

technical steel research

Physical metallurgy and design of new generic steel grades

Constitutive modelling for complex loading in metal forming processes

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Abstract

The key objectives of this project were:

1. To investigate, by laboratory test simulation and hot rolling trials, the effect on the microstructural evolution of the loading path taken to achieve a given overall cumulative strain.
2. To identify the mechanisms responsible for differences in constitutive behaviour.
3. To formulate constitutive models which help provide an understanding of this behaviour. Instances of constitutive models are obtained for a range of steel qualities, including austenitic, ferritic and duplex structures, using experimental data obtained from the laboratory tests, are validated using results from such tests and hot rolling operations to which the models are applied.

To achieve these objectives the partners have cooperated in supplying materials, in using their experimental testing and rolling equipment to implement the required loading paths, in providing metallurgical services for material characterisation, and in sharing their facilities for development and validation of the constitutive models.

It has been found, in this work, that comparable mechanical loading paths in the experimental laboratory test simulation trials have given rise to significant differences in rheological and structural behaviour:

1. An abrupt decrease in strain-rate, in tension tests can give rise to a lag in the flow stress behaviour behind that for a mechanical equation of state, and an even greater lag in the rate of static recrystallisation, and, in the case of ferrite, cells larger than those obtained without the drop in strain rate.
2. Strain reversal in both tension/compression and torsion tests produces a noticeable Bauschinger effect and a plateau in the flow stress curve. For interpass times greater than 10 s, there is also a change in the static recrystallisation rate depending on the magnitude of the strains applied. For ferrite, the grains are coarser, the cells larger and low angle boundaries more numerous.
3. A strain reversal coupled with an abrupt decrease in strain rate gives rise, for both torsion and tension/compression, to a Bauschinger effect and plateau more noticeable than without the drop in strain rate. Conversely, if the change in strain rate is an increase, results for torsion exhibit opposite i.e. less pronounced results.
4. A double hit in the same direction can cause a lag in the recrystallisation curve behind that for the first compression but, if the reductions are sufficiently high, the curve will ultimately be higher than that for a single reduction of the same

total absolute strain. Equivalent triple hits with individually lower reductions may, however inhibit recrystallisation completely. If the two compressions are applied in perpendicular directions, recrystallisation may take longer to complete, although in the short term, the rate may be faster.

For comparable thermal loading paths in an austenitic steel, differences are insignificant.

In the laboratory rolling trials, differences have been observed between loads for the high and moderate speed rod rolling schedules and between loads for the monotonic and forward-reverse plate rolling schedules. In the latter application, differences in deformation have also been found. However, unlike experimental tests, effects of loading path on structural evolution, if existent, appear at best to be secondary. Difficulties in detecting any differences have arisen because of the presence of phases such as bainite and wide variations in grain size associated partly with heterogeneity in the initial structure. Equally importantly, the angle between corresponding strain paths is not large, overall, for the simple rolling processes considered.

To predict the effect of loading path on material behaviour, constitutive models have been formulated by all the partners. In some cases, this has been achieved by incorporating observations from rolling trials or experimental tests, as in the viscoplastic model of TU Freiberg used to examine the effect of reversal on deformation in plate rolling and in the empirical model of CEIT for predicting the complete stress-strain curve associated with strain reversal. Other models are more phenomenological in nature, in which the key mechanisms of dislocation dynamics have been incorporated as in the analytical models of the University of Oulu for predicting the stress-strain curve associated with abrupt changes in strain-rate and of CEIT for the simulation of recrystallisation following strain reversal. Generally, these models are numerical, requiring implementation using FE, such as the those used by Corus and MEFOS for predicting the recrystallisation kinetics and evolution of the grain size and dislocation distribution for the H-V rolling trials and the benchmark application viz. the rolling of CMnNb steel plates at TU Freiberg.

An alternative approach, not including dislocation density, has been followed by CSM in the formulation of a numerical, multilevel model for the simulation of hardening and dynamic recovery, dynamic recrystallisation, and static softening. The model has been set up and validated for a fast recrystallising medium C steel and a non-recrystallising duplex stainless steel using experimental flow stress data for tension/compression and torsion from the University of Oulu and CEIT. The model has been validated for laboratory plate and industrial bar rolling and successfully used in the study of shear reversal.

For all the models good agreement has generally been obtained between predictions of deformation and load, and experimental measurements and, where the latter can be obtained, of metallurgical structure.

In conclusion, experimental tests have clearly demonstrated that the flow stress and microstructure can be significantly different for various loading paths used to achieve a given state of equivalent strain, depending on the magnitude of the angle between the paths and of changes in strain rate. Constitutive models have also been formulated and successfully tested and have the potential, in conjunction with appropriate laboratory trials and accurate techniques for material characterisation, for investigating the effects on microstructure, of a wider variation of strain paths in forming processes more complex than those considered hitherto.

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1. Introduction

1.1 Complex loading

In the context of process history in hot rolling, loading can be either mechanical or thermal. The mechanical loading usually refers to the roll separating force but is also associated with front/back tensions and loads accompanying imposed boundary conditions. Thermal loading is associated with reheating and cooling and should be included in this context.

Complex loading, by definition, occurs when there are rapid or abrupt changes in the loading path. During deformation, it can thus arise, for example, from rapid changes in strain rate and temperature, shear strain reversal, microstructural changes associated with dynamic recrystallisation etc and material inhomogeneities causing abrupt changes in stress levels as the stock progresses through the roll gap. Complex loading needs ideally to be considered in the context of the complete process route, but in the project is studied for at most 3 pass schedules including interpass cooling and reheating. In this case, complex loading is associated with changes in the deformation path such as interpass turns of the stock in H-V rolling and reversing of the stock or rolling direction in plate rolling. Load changes also arise as a result of static recrystallisation , cooling etc in the interpass.

It has been demonstrated in the final report of a previous ECSC project [1] (the forerunner to the current project) that different yield stresses may be obtained if different loading paths are used to obtain the same final dimensions. More generally, it has been shown that for certain metals, the thermomechanical history and microstructural evolution may significantly affect the properties and influence the magnitude of the loads and other operating parameters during processing [2]. It is therefore important to establish, for which cases i.e. forming processes and steel qualities, the forming route is significant.

Significant effects are more likely to be observed when the path changes abruptly for instance if there is a substantial change in the strain angle, Θ , defined by the equation

$$\cos \Theta = \frac{\varepsilon^1 : \varepsilon^2}{\|\varepsilon^1\| \|\varepsilon^2\|} \quad \dots 1$$

where ε^1 , ε^2 are the strain tensors at the point considered before and after the change in strain path and represented relative, to a coordinate frame, in the conventional manner, as a 3x3 array with normal strains along the diagonal and shear strains off the diagonal.

Application of the above formula assists in choosing deformation processes in which there is a significant change in strain angle. It also indicates that a complete reversal in strain, or even changes of 90° or more, may be difficult to realise in practice. Indeed, in phenomena such as shear reversal, although significant curvature in the strain path seems plausible, the change in strain angle is actually of the order of 60°. The changes likely to be observed in practice thus depend on the metal forming processes of commercial interest and on the experimental facilities available. The Department of Engineering Materials at Sheffield

University has already carried out valuable work on the influence of strain path on metallurgical structure [3] using an ASP (arbitrary strain path) machine as well as a Gleeble machine and close liaison has been maintained through the IMPPETUS program [4].

1.2 Constitutive modelling

In order to understand the relationship between microstructural evolution and process history for such complex loading paths, experimental testing alone is important but insufficient; it is essential also to predict, with sufficient accuracy, both the process parameters and microstructural characteristics for current and novel forming schedules. Consequently, an appropriate mathematical model needs to be selected and tested, if available. Therefore a review needs to be carried out of models by the partners, of models described in the literature and modelling facilities of third parties.

Many models, as to be expected, are available or potentially developable. These may be classified in several ways, each type having its potential uses and drawbacks [5]. At one extreme are the empirical models obtained by fitting curves to experimental data without consideration of the reason, if it exists, for the relationships between the variables, and similarly, neural network models. At the other extreme are the phenomenological models based on 'fundamental' laws governing natural phenomena. The material properties in this case consist of 'fundamental' experimental data. Phenomenological models, in principle, can be applied to new problems with a minimum of new data i.e. sufficient to define the problem. In practice, the models may be very complex so that analytical solutions are rarely possible and so require numerical solutions obtained using finite elements etc. Generally, the models employed have been deterministic, but if it is difficult or impossible to observe a quantity, e.g. grain size then such a quantity can be replaced by a suitable statistical distribution; such models are termed probabilistic or stochastic if relating to temporal evolution. Generally, simplicity rather than complexity is required. One may wish to take on board the physical laws to predict the general behaviour without handling the difficult equations. This can sometimes be achieved by qualitative physical models.

Models can also be classified according to the scale or levels at which they operate, different but often relatable physical 'laws' being used at each level [6]. The number of levels used and their definition in the literature varies somewhat, but broadly consist of (a) the macroscopic level, at which the operating parameters can be related to reduction and bulk temperature etc but not to the structural parameters (b) the mesoscopic level, at which the standard models for internal variables such as stress and strain, obtained by averaging over a length scale of c. 0.01–1.0 mm, are formulated (c) the 'microscopic' level, depending on whether optical or electron microscopy is used, dealing with evolution of grains and dynamics of dislocations. Between (b) and (c) a number of 'mezza' levels have sometimes usefully been introduced, allowing the gradual introduction of quantities such as hardness, drag stress, fraction recrystallised etc which reflect selected aspects of the microstructural behaviour. Also below (c), levels have been introduced, the 'submicroscopic' being relevant in this context and dealing with nucleation of grains from subgrains, or, in modelling damage, the migration of atomic vacancies to form voids. In reviewing and formulating constitutive models, attention has been focused on levels (b) and (c).

Constitutive equations describe the response of the body to applied loads resulting from the internal constitution of the material and so in this context include the various equations of elasticity, plasticity and heat transfer. A theory of constitutive behaviour, in which the constitutive equations are required to satisfy certain axioms and from which various types of constitutive behaviour/equations are classified, was formulated by Truesdell and Noll and Coleman [7,8]. The application of this theory to continuum mechanics is described in detail

in the book by Malvern [9] and the potential application to metal forming in particular described in a previous ECSC project [10]. Of particular relevance to complex loading, hereditary behaviour is described by equations in which the stress is a functional of the previous history and, of possible relevance to rolling texture, simple materials are defined in which the stress is dependent on the history only of the deformation gradient. For some materials, viz. those with 'fading memory' only the recent history has a significant influence. In metal deformation, experiments have sometimes detected a slight lag in this response whereas in classical models of plastic deformation as used in rolling it is assumed that there is an immediate stress response to an applied displacement i.e. that the material satisfies an equation of state. These observations suggest that, to investigate the effect of deformation history, models along the above lines should be developed.

The equations referred to above are very difficult to solve efficiently and it is recommended that solutions be deferred. In the short term alternative ways can be found of incorporating process history the results of which can ultimately be compared with the those of the more complex equations and with experimental results, allowing the accuracy and efficiency of each solution approach to be assessed. One method is to update the programs with information on the deformation etc. gleaned from experimental results showing a history effect. It should also be noted that the equations do not directly predict the effects on metallurgical structure.

Not only is the metallurgical structure of a formed product a response of commercial interest in the project, it is a signature of the physical state of the latter and therefore a key factor in identifying, and understanding the effects of, process history. It is possible that the metallurgical structure of two products with the same external dimensions will be different, if they are formed by different process routes.

An alternative approach may therefore be adopted based on the assumption that there are state variables, related to characteristics of the microstructure, which encapsulate these history effects. Understanding the evolution of metallurgical structure may be achieved by means of constitutive equations describing the underlying microscopic behaviour of the material, in terms of the coupling between such structural variables and the thermomechanical variables, both during and between deformations. Several constitutive models are available, depending on the mesoscopic model adopted for the thermomechanical variables e.g. (elasto)(visco)plastic, small or finite strain [11,12], the accompanying structural variables [13] and the underlying assumptions based on thermodynamics [14,15]. Comparative reviews of constitutive theories in cyclic plasticity have been given by Chan [16] and a unified treatment of the various viscoplastic theories given by Miller [17] and contained in the book by Krausz & Krausz [18]. Such models can be implemented on the computer via a range of FE codes (see Refs. [19,20]).

1.3 Experimental investigation of constitutive behaviour and material characterisation

Development of such constitutive equations entails the carrying out of experimental trials to obtain data from which the coefficients can be extracted, to help validate them once formulated and to examine the influence of loading path on structural evolution, in particular recrystallisation. Various experimental techniques are used to carry out this examination and have been reviewed by Dzubinsky [21]. Some are mechanical in nature using stress relaxation or double hit tests and others are direct observation methods using either optical or electron microscopy (e.g. EBSD/SEM, TEM).

The basis of the stress–relaxation (SR) test [22,23] is that the decrease in stress following deformation can be divided into 3 discrete stages consisting of creep, recrystallisation and stress relaxation of soft material. An equation has been proposed [24] enabling the fraction recrystallised to be ascertained from the characteristics of the curve, however stress remaining during the test can sometimes promote climbing of unstable dislocations and thereby accelerate recovery. Stress relaxation tests can thus prove insensitive so double hit tests are used instead in such cases. In the double hit test (DH), the fraction recrystallised is calculated by finding the changes induced in the second deformation; this is achieved either by back extrapolation, comparing areas or by comparing the difference between the mean yield and an offset yield stress for both hits. It is generally less reliable than stress relaxation when the latter can be employed.

In optical metallography (OM), the fraction recrystallised is found in section(s) of a suitably prepared sample either by finding the ratio of the total area of regions recrystallised to the area observed, by similarly comparing intercepts on a given line with the total length or by finding the fractional number of a points in a random set that fall within the recrystallised region. The method is relatively simple to apply, although sometimes time consuming, but significantly subjective. It is difficult to apply for low carbon steels and when observing complex microstructures that have been austenite quenched. Austenite structures can be difficult to detect when transformation has occurred and bainite or pearlite have been formed; they are easier to observe when martensite has been formed but this generally requires a rapid cooling rate and thin sections. It can also be difficult to distinguish between recrystallised and deformed grains particularly if the sample has been subjected to a compression that was axisymmetric. OM is also used to measure grain sizes and to describe the local distribution (usually assumed to be lognormal) by calculating the average grain size and standard deviation. If there is not a wide variation in grain size, and the original grain sizes are equiaxed then calculation of the major and minor axes of the deformed grain can be employed to calculate the strain distribution using the Hensler-Gifkins method [25]; the strains can then be used in model validation.

Of the techniques considered, EBSD mapping is potentially the most accurate. The electron backscatter patterns of Kikuchi bands produced by diffraction of a beam from a Scanning Electron Microscope, on striking the crystalline sample, are used to obtain Orientation Imaging Micrographs which provide a visual characterisation of the orientation characteristics of the microstructure. By combining these micrographs with a band contrast map, a measure of the sharpness of the EBSD pattern above the background is obtained, which, in principle, enables recrystallised and non-recrystallised regions to be identified. A number of micrographs are employed e.g. image quality maps, orientation spread maps and inverse pole figure maps. The former describes the quality of the pattern and provides a qualitative indication of the microstrain distribution, a high IQ value suggesting recrystallisation. The orientation spread for each grain is obtained by averaging, over points of the grain, the absolute deviation from the mean orientation of the grain; a small value indicates recrystallisation. Inverse Pole Figure Maps use colour to represent the orientation, a predominance of white suggesting little likelihood of recrystallisation. For maximum accuracy, the maps should be used in combination. The EBSD/SEM method, however, does have some limitations (a) as for OM, it can only determine the fraction recrystallised if no phase transformation has occurred after recrystallisation and does not easily differentiate between deformed and recrystallised grains, particularly if the sample has undergone axisymmetric compression, (b) the accuracy depends on the equipment used, a tungsten filament being less accurate than a field emission system.

2. Project details

2.1 Objectives

The objectives as extracted from the Technical Annexe (included in Appendix 8) are, in brief:-

1. To investigate the constitutive behaviour of steels during complex loading paths
 - (a) by laboratory test simulation
 - (b) by laboratory hot rolling trials
2. To identify the mechanisms responsible for the constitutive behaviour.
3. To formulate constitutive models, including these mechanisms, for predicting flow stress, microstructural evolution, etc.
4. To incorporate the constitutive models into FE models of rolling and validate using full scale hot rolling.

2.2 Activities

Outlining in more detail the activities associated with the above objectives, the project investigates the effect of loading path for hot rolling processes of commercial interest to the partners and within a given remit (avoiding processes of very high complexity) viz. the rolling of bar, rod and plate. A wide range of steel qualities of commercial interest to the partners are studied viz. austenitic, ferritic and duplex, with fast and slow recrystallisation kinetics, transforming and non-transforming and in both cast and wrought conditions. It was also planned, however, that all partners should study similar steels viz. CMnNb and stainless steel.

The complex loading paths considered, as stated at the outset are strain reversal and those related to the interpass behaviour i.e. a change in stock orientation in the three principal planes and reverse deformation in tension/compression and torsion. Accordingly standard experimental testing facilities are employed viz. hot torsion and Gleeble machines and also laboratory rolling mills.

The effect of various loading paths on operating parameters and microstructural evolution are assessed experimentally, the microstructural aspects of main interest being grain size, fraction recrystallised and dislocation density. Material characterisation is carried out using both direct and mechanical methods as indicated in Section 2.3 viz. optical metallography, electron microscopy, stress relaxation and double hit techniques.

In order to understand the effect of loading on structural evolution, constitutive models are formulated and validated using experimental data and applied to laboratory and industrial hot rolling processes.

2.3 Partner resources and interactions

A breakdown of the project activities and the linkage with the technical resources and areas of expertise of the partners is given in Table 1 and shown diagrammatically in the tree diagram (Fig. 1) constructed using 'Mind Map' software [26]. From scrutiny of this figure, and using network analysis etc. the commonalities of interests, requirements and

availabilities or otherwise of facilities to meet these requirements can be seen and are the drivers for the partner interactions. This information has been extracted and presented in Table 2. The distribution of these amongst the partners is indicated in a matrix Table 3 and summarised below.

- (a) Materials; A range have been used:- Armco Iron and several steel qualities viz. low and medium C, CMnS, CMnNb, CMnNb hiTi, CCrV IFNb. IFNb Ti, stainless steels (Type 316 and duplex). Of these CMnNb, medium C and stainless steel (duplex) are of widespread interest to the partners, in either experimental or theoretical simulations and have been chosen as benchmark materials.
- (b) Experimental Testing Facilities: All partners have access to experimental testing facilities, e.g. compression machines at the University of Oulu and Corus, torsion machines at CEIT and TU Freiberg and laboratory rolling mills at TU Freiberg, CSM and Corus.
- (c) Pilot/Industrial Rolling Mills accessible/owned by CSM and Corus.
- (d) Metallurgical facilities: all partners employ metallographic or optical microscopic techniques for characterisation of structures.
- (e) Mathematical Models: All partners have developed or have access to physical, phenomenological or empirical models for describing both forming processes and experimental data. Since they also have a common interest in recrystallisation, either static or dynamic, this phenomenon is covered in all models. Other phenomena are represented in some of the models e.g. recovery, dislocation dynamics, hardening, vacancy generation etc.
- (f) Finite Element Software: Most of the partners are using FE packages (e.g. ABAQUS or LS Dyna) for incorporation and implementation of the above models.

Partners have collaborated in providing particular resources where these are needed or in sharing information where there is a common interest.

Discussion has also taken place between the partners on how to present their activities in a common framework. It was agreed that this framework should indicate the experiments or models used for various steel qualities and the range of temperatures, strains and strain-rates employed. It would thus indicate where comparisons may reasonably be made between (a) model predictions/experimental results and (b) steel properties. Equally it would show where further data were required to complete the assessment.

Details have been compiled from communications with partners and from the contents of previous technical reports and are shown in Table 4(a-f).

Key information giving an overall picture of the work carried out by the partners has been extracted from these tables and has been input to a template designed at the University of Oulu. This completed template is included in Appendix 7.

The progress in this period made by each of the six partners towards achieving the above objectives together with plans for further work is summarised in the section below and described in detail in Appendices 1 to 6.

2.4 Structure of project

In accordance with the objectives stated in the Technical Annexe and as described therein, the project is structured into tasks or work packages carried out by subsets of the partnership and further divided into subtasks as indicated below: -

WP1: Evaluation of Constitutive Models (Corus, MEFOS, CSM; CEIT, the University of Oulu, Freiberg)

WP2: Experimental Investigation of Constitutive Behaviour (the University of Oulu, CEIT; Freiberg)

- (i) Strain Rate and Temperature Changes (the University of Oulu)
- (ii) Strain Reversal in Tension/Compression (the University of Oulu, Corus, CEIT)
- (iii) Strain Reversal in Torsion (CEIT)
- (iv) Multi-directional Strain Tests (Corus UK)
- (v) Ferritic Tests (the University of Oulu)
- (vi) Stock Rotation during Rolling (Corus, Freiberg)
- (vii) High Speed Rolling (Freiberg; Corus)

Note: The original list of subheadings has been rationalised by the combination of topics 2.2.6, 2.2.7 in the Technical Annexe as subtask (vi) since they both relate to stock rotation during the interpass and, in practice, involve the same partners.

WP3: Material Characterisation (CEIT, the University of Oulu; Corus)

WP4: Formulation of Constitutive Laws (Freiberg, MEFOS, CSM, Corus; CEIT, the University of Oulu,)

WP5: Incorporation of Constitutive Laws into FE Models of Rolling (MEFOS, CSM; Freiberg, Corus)

WP6: Model Validation (CSM; MEFOS, Corus, TU Freiberg)

3. Summary of progress (Details given in Appendices 1 to 6)

3.1 WP1 Review of Constitutive Equations

All the partners have been working in previous ECSC projects involving the modelling of metal forming and have brought with them a legacy of models of various types, as mentioned in Section 1.2, which have been reviewed for either immediate use in the project or further development in describing the influence of strain path on the evolution of microstructure and on the magnitudes of the operating parameters. The literature has also been reviewed for current developments in modelling and contact made with third party developers of software where relevant.

The models can be divided into two groups viz. (a) empirical or analytical formulae for (i) the operating parameters in terms of reduction, bulk temperature, average strain rate (or speed), yield stress and friction and (ii) metallurgical parameters such as fraction recrystallised in terms of similar quantities to (i) but ostensibly at the mesoscopic scale and (b) finite element plasticity models for the prediction of the distribution of these quantities based on standard constitutive equations at this scale. Both these groups of models provide means for investigation of influence of thermomechanical history, which are straightforward but have limitations. In the first group, determination of any history effect generally requires taking some accompanying measurement(s) in a corresponding physical trial. For example

1. Load models require the mean forming resistance over the strain path, which is dependent on the microstructural behaviour but is estimated from experimental data using procedures such as those set up by TU Freiberg.
2. The probabilistic model used at the University of Oulu for the time to 50% recrystallisation in terms of mean stored energy expresses the latter in terms of the dislocation density. The dislocation density is not determined by recourse to an evolutionary model but estimated empirically in terms of the measured flow stress. Indeed this approach accompanied by appropriate experiments have readily demonstrated that the recrystallisation is not strongly dependent on the temperature history but determined essentially by the temperature prevailing in the interpass.
3. CEIT have viscoplastic and dislocation density based constitutive models, both largely analytical in nature. The first model takes into account the effect of strain reversal on the flow stress curve. However analytical modelling of the transient after reversal is difficult and the contribution of non-recrystallised regions needs to be determined experimentally and is strongly dependent on the dynamically recrystallised fraction, which cannot be precisely determined in terms of a macroscopic external measure.

The intrinsic nature of the plasticity equations implies that the stress and strain distributions in the stock at any stage of the process are dependent in general on the deformation path taken to reach this stage where path is taken to mean a sequence of reductions and interpass turns. FE codes such as ABAQUS used by Corus and CSM can then determine the effect of this sequence on the mechanical state of the stock, and also on structural quantities such as dynamically recrystallised fraction and grain size (having checked that the critical strain has been attained) by means of user defined variables. Such codes do not however take into account the coupling between microstructural and thermomechanical quantities. Equations coupling these quantities can be introduced into ABAQUS using the *CREEP Option, provided that kinematic hardening is excluded, and without this exclusion by means of a user subroutine (i.e. UMAT for ABAQUS/Standard and VUMAT for ABAQUS/Explicit). Corus has developed an FE model, by means of the latter approach, incorporating kinematic and isotropic hardening; the numerical procedure for time integration of the constitutive equations may, however, converge very slowly or even diverge. An FE model taking into account the interaction between the mechanical variables, hardening, recovery, recrystallisation, and damage has also been developed by CSM in the previous project. It has been used to investigate strain reversal, in particular, the different material responses noticed by CEIT for shear and normal deformation. More work was necessary however to study the latter effects in rolling and to investigate the complex microstructural changes for both single phase and duplex structures. The interaction between deformation and dislocation behaviour has been studied in the previous project by MEFOS using the Sandstrom-Lagneborg, Engberg-Siwecki and Bergstrom models relating flow stress,

recrystallisation, dislocation and vacancy concentration including evolutionary equations for the latter two quantities and has been applied to the rolling of CMn and CMnNb steels in a single pass. The model has potential for further development and application to rolling.

From this review, the following recommendations were made for the FE models:

1. Augmentation of finite element models to include all processes of interest such as heat transfer, static softening, etc.
2. Utilisation, where possible, of fundamental models such as those based on dislocation density.
3. Incorporation in dislocation models of relevant processes such as nucleation and precipitation.
4. Use of efficient numerical techniques for time integration.
5. Determination of coefficients for all steel qualities allocated.
6. Development/application of models for multipass rolling including effects of cooling especially in the interpass.

Concerning the application of analytical and empirical models, it was recommended that supporting measurements be replaced by a constitutive model where possible. It is also important that the skills and expertise gained in measurement techniques be used to full advantage (a) to provide constitutive data for use in FE models both locally and by the other partners and (b) to investigate experimentally the effect of a range of complex loading paths on structural evolution.

3.2 WP2 Experimental Investigation of Constitutive Behaviour

(i) Strain rate and temperature changes

Experimental work has been carried out at the University of Oulu to investigate the influence of deformation history, in particular, changes in temperature and strain rate, during uniaxial compression/tension on the flow stress and subsequent softening behaviour in both the hot-deformed austenite and warm-deformed ferrite phase (see Section 2.2 (v)).

It was found that when the strain rate decreases abruptly, in the dynamic recovery regime, the flow stress lags behind the behaviour of a mechanical equation of state, so that it is finally higher than that at the corresponding constant final strain rate, close to the stress at the constant mean strain rate of the whole deformation history. The static recrystallisation rate lags even more than the flow stress from the behaviour following the mechanical equation of state, so that following the abrupt decrease in the softening rate remains close to that at the high strain rate before the drop. Since the effect of strain rate on recrystallisation kinetics is generally weak, the mean strain rate can be used for modelling with reasonable accuracy. The recrystallisation behaviour, however, is affected quite significantly in the instances that the strain rate drop takes place after a strain resulting in dynamic recrystallisation. The above effects have been shown to be independent of the steel composition.

Similar work has also been carried out by Corus for compression and by CEIT in torsion tests.

The influence of temperature increase was investigated only for the austenite phase. The results generally indicated that the temperature history seems to have only an insignificant influence on the flow stress. Similarly its effect on subsequent softening behaviour is minimal.

(ii) Strain Reversal in Tension/Compression

The influence on flow stress and static recrystallisation kinetics of strain reversal in axisymmetric deformation has been investigated by the University of Oulu and Corus for CMnNb steels supplied by Corus and various other steel qualities. Tests by all partners have shown that strain reversal gives rise to a noticeable Bauschinger effect, a transient characterised by a plateau. The recrystallisation kinetics is strongly retarded by the reversal, especially at small prestrains. This effect seems to be quite independent of the chemical composition of the steel. A strain reversal coupled with a drop in strain-rate is found to decrease the SRX rate even more. Contrary to this, partial recrystallisation after the first deformation before reversing the strain enhances slightly the recrystallisation. CEIT have shown that strain reversal in torsion coupled with an increase in strain rate tends to increase the stress after the change and to shorten or remove the plateau. The effects from the above two coupled tests can be related to an excess or shortage of dislocations in relation to those due to the dissolution of the substructure.

(iii) Strain Reversal in Torsion

Torsion tests have been carried out on CMnNb, medium C and duplex stainless steels in order to investigate the effect of strain reversal on the flow stress under a broad range of conditions and on the recrystallisation kinetics.

For CMnNb steels, full – or even relatively large partial –reversal gives recrystallisation times shorter than those for the pre-strain only but longer than if the total absolute strain is applied monotonically. For small reverse strains, the recrystallisation time is approximately the same, or even longer if the pre-strain is low i.e. c. 0.15, the longest times occurring for reverse strains close to the end of the plateau.

For medium C steel, tests have been carried out to simulate the sequence of shear reversals taking place during a rolling schedule assuming no softening between passes and to obtain constitutive experimental data for the CSM FE programs (pure shear). It has been notably observed that the flow stress is lower in a 2nd pass with a prior reversal than without. This effect has also been observed in the non-recrystallising duplex stainless steel. Multipass tests have also been carried out to determine the softening taking place during given holding times applied after different strain paths. For forward and reverse strains of 0.1, interpass times greater than 10 s are required to initiate static softening, but the softening is significantly increased if a higher strain of 0.2 is used. A series of cyclic tests of small strain produce stresses in successive passes significantly lower than those in a monotonic test having the same equivalent strain.

(iv) Multidirectional Compression Tests

The objective of this subtask was to investigate the effects of successive hits of various magnitudes applied isothermally, but varying the loading direction on the flow stress and recrystallisation kinetics of Type 316 stainless steel. In particular it was aimed at comparing the effects of a single hit and a series of hits of the same total strain whether applied on the same or on different pairs of faces (i.e. indifferent directions relative to axes in the specimen).

The order of the hits and the operating temperatures viz. 900°C, 1000°C and 1100°C were also examined. To maintain isothermal conditions, specimens were quenched and reheated between hits.

The changes of orientation of the specimen relative to the platens in the main reflect the configurational changes occurring in industrial rolling processes such as turning of a bar during edge rolling and cross rolling of plates. More general changes e.g. turning of the whole stock in more than 1 plane, although not representative of bulk processes in conventional rolling, may nevertheless be applicable locally to regions of the stock undergoing complex deformation e.g. near the web flange interface in beam rolling.

Stress relaxation tests carried out to measure the recrystallisation indicate that

1. The time for recrystallisation in a single hit decreases notably with strain and marginally with increasing temperature.
2. Recrystallisation behaviour is marginally dependent on the order of the hits.
3. Multiple hits of the same total magnitude as a single hit give a lower recrystallisation and, triple hits each of magnitude (~20%) applied in various directions can inhibit recrystallisation entirely.
4. An interpass turn of 90° enhances the early recrystallisation when one of the reductions is large (~40%) but significantly retards it when both hits are moderate (~20%).

(v) Ferritic tests

The influence of strain reversal and strain-rate change on flow stress behaviour and evolution of recrystallisation has been investigated by the University of Oulu also for the ferrite phase, using compression testing and optical and electron microscopes. This information may be especially useful in the development of future models for the ferritic rolling. As the softening of ferrite is affected strongly by the recovery, stress relaxation techniques could not be used to reveal the recrystallisation rate, but the double-hit compression technique had to be applied. The influence of strain rate change has been studied for the various IF steels and hi Ti-Nb steels. The results obtained are qualitatively similar to those in the austenite phase. When the strain-rate decreases abruptly, the flow stress and, even more noticeably, the static recrystallisation rate lag behind those of a mechanical equation of state. After the change, the cells appear somewhat larger than when the high strain rate is maintained indicating a rapid decrease in the driving force for static recrystallisation. The misorientation distribution of grain boundaries is, however, unaffected. In particular for small reversed strains, the strain reversal retards the recrystallisation rate of ferrite – by as much as 8 times – and significantly coarsens the ferrite grain size and the cell structure. It also tends to decrease the average misorientation across the grain boundaries. Strain rate tests have also been carried out on 55Cr3 steel to obtain constitutive data in co-operation with CSM for use in its rolling model. Tests have also been modeled using the model developed at MEFOS to understand the relationship between flow stress changes and dislocation dynamics (see Section WP4).

(vi) Stock rotation during rolling

(a) H-H/H-V rolling

Two series of trials using Type 316 stainless steel bars were carried out on the local Cavendish mill, the first to compare the effects on structure when rolling with and without an interpass turn of 90° between 2 successive 20% passes at an initial temperature of ~1100°C, and the second to investigate the effects of the above when combined with various reheating strategies in the interpass.

1. In the first set of trials, lower recrystallisation was observed at the surface but the loads in the second pass are lower suggesting that retardation of the recrystallisation is localised.
2. In the second set of trials, four bars were rolled in the first pass at a lower temperature than in the first set of trials, viz. 1040°C. The bars were then quenched and two reheated to 1040°C to promote recrystallisation and two to 950°C to inhibit it. One bar in each of these sets was rolled in the second pass without a prior turn. The other bars were turned respectively in transverse and horizontal planes.

Conclusive evidence cannot be given on the effect on recrystallisation since determination of the latter by optical metallography has been considered unreliable and also the grain sizes observed varied widely, especially for the bar reheated to 950°C and not turned. The fact that the loads in the second pass were similar for both reheating temperatures suggests that little recrystallisation has been promoted. This is also suggested, albeit weakly, from pole figure maps obtained by EBSD.

The set of grain size results give an overall view of the maximum rather than the average grain sizes attained and suggest that the maximum is greater in the specimens that have been turned, the substantial differences are however more likely to reflect material heterogeneity than a retardation in grain refinement.

The hypothesis made following these tests is that any effects of the stock rotations and reheating strategies on recrystallisation in a rolling context are secondary.

(b) F-R Rolling

The objectives of this work were (a) to investigate, for various rolling temperatures, the effect of an interpass turn through 180° on the recrystallisation kinetics and evolution of austenitic grain structure in CMnNb steel plates (b) to perform a benchmark test showing the effect of stress-strain history on the rolling force and microstructural evolution. Rolling trials were carried out both on the plate mill at Swinden Technology Centre and the reversing stand mill at TU Freiberg.

Plate rolling schedules, with and without reversal, were first designed and implemented at STC, in conjunction with FE and metallurgical modelling, to guide the plate rolling trials at Freiberg. The rolling temperatures employed were 1020°C and 1050°C and the plates were reduced in thickness from 30 mm in 3 passes of draughts 5 mm, 3 mm and 3 mm with and without reversal. Measurements of loads and spread were in good agreement with values predicted by the models. However, although large grain sizes were predicted by the metallurgical model, the austenitic grain structure could not always be differentiated because

of the presence of phases such as bainite, and decarburisation at the surface and where the presence of martensite did allow this structure to be detected, the grain sizes varied widely.

In the schedules at Freiberg, plates were employed of thickness 6.7 mm, this having been calculated as the maximum value for which the cooling rate would be sufficiently high to produce martensite. However, trials were also carried out using the 30 mm thick plates, as supplied by Corus, the temperature being more stable and controllable than in the plates of reduced thickness. The stock was reduced in thickness, firstly by 40% and then by a further 50%, with and without reversal between the passes, the plates being moved manually to the roll bite in the first case, and turned through 180° in the second. Rolling temperatures of 1000°C, 900°C and 850°C were employed. Small differences were observed between the dimensionless forming resistance (and hence the loads) with and without turns for the 6.7 mm thick plate, it being higher in the former case at 900°C but lower at 1000°C (undecidable at 850°C because of the different L/h values actually used).

As in the trials at STC, decarburisation occurred and no martensite was produced in the thicker plate but was produced in the initially 6.7 mm thick plate. Inhomogeneous grain size distributions were obtained, the grain size tending to be smaller in the plates that had undergone reversal rolling, supporting the hypothesis that the different strain pass histories give rise to different average grain sizes. A combination of viscoplastic techniques, analytical and FE models (as described in Sections 2.4 and 2.5 for tasks WP4 and WP5) have been used to obtain predictions which can be compared with experimental results. The FE results obtained by Corus and MEFOS for this benchmark application are discussed in Section 3.6 for task WP6.

(vii) High speed rolling

The objective of this task is to compare the development of austenitic grain size in CMnNb steel at low speeds with that at high speeds such as occur in rod rolling. TU Freiberg has collaborated with Corus in the use of its high speed rod mill (top speed ~ 32 m/s) and also in the use of its plate mill (as described above) in providing results at the lower end of the speed range (~1 m/s). Moderately low speeds of the order of 5 m/s, attainable on the rod mill were also employed. A constant roll pass design (round – oval; 10 mm → 6 mm) was employed for all trials except those at low speed. In high speed rolling, reheating temperatures of 1200°C and 1150°C were employed, for times of 10 s and 60 s (approximately). All stands F1 –F4 were used. Air cooling was used, At low speed speeds, trials were carried out using all stands F1 –F4 and water cooling, F1-F2 and water bath and F1 only with the water bath.

Information from Corus at STC on the dependence of recrystallisation time on temperature, initial grain size (in this case 50 microns) and strain was employed in assessing these trials.

The experimental results, however, show only slight differences in grain structure for the various cases considered viz. higher rolling speeds produce smaller grain sizes when used in conjunction with higher input of thermal energy (i.e. higher reheating temperatures or reheating times) and so are comparable with those produced by reversal at the lower speeds in plate rolling. Notable differences in load are found viz. higher values at higher temperatures (at constant speed), probably arising from changes in material behaviour associated with phase transformation.

3.3 WP3: Material characterisation

Information on the materials examined by the various partners and the corresponding metallurgical processes studied has been input into a template which has been involved in the testing of most of the metals. The information is reproduced in Appendix 7. Details of the materials and associated tests and models for the individual partners are provided in Table 4(a-f).

The results from the various testing facilities of the partners for the materials listed in these tables are outlined below.

1. CMnNb steel is produced by Corus and it has been decided that this should be one of the major steels used by several partners for comparison purposes. Accordingly, it has been supplied to TU Freiberg for benchmark plate rolling trials and to CEIT for torsion testing. Steels of a similar composition and CMnNbTi steels, have also been used by the University of Oulu in its compression/tension testing programme. It has been used by Corus in this project in strain reversal tests and plate rolling trials.

The torsion testing has found that the flow stress stagnates following reversal, probably due to strain induced precipitation in the niobium-carbonitrides. Most significantly, it transforms in the range 750°C to 800°C, in a manner dictated by the cooling rate. If the latter is sufficiently high, corresponding to a time $t_{8/5}$ of less than 1.3 s martensite is produced which is beneficial to the detection of austenite, otherwise phases such as bainite and pearlite are produced which mask the austenitic structure.

2. CMn steels (i.e. without the addition of Niobium) have been used to a limited extent by Corus, having been extensively used in the previous project. In particular, the dependence of recrystallisation on parameters such as grain size and interpass time was sought to help determine an appropriate schedule for the rolling trials on the rod mill at TU Freiberg. In particular, it was found that coarsening the initial grain size significantly increases the recrystallisation time making partial recrystallisation possible before entry to the next pass.
3. CMn steels with the addition of sulphur have also been tested by the University of Oulu and characteristically have an insignificant grain size evolution below 1200°C.
4. The medium C steel 55Cr3 has also been investigated by several of the partners, being used in the rolling trials on the ORI-MARTIN bar mill, modelled by CSM and supplied to CEIT and the University of Oulu for obtaining torsion and tensile/compressive constitutive data and for validation purposes. It is characterised by rapid recrystallisation and grain growth, thereby wiping out strain history effects in the bar rolling process.

The evolution of grain size in a medium C steel has also been studied by Corus to assist the heat treatment process at TU Freiberg in achieving an initial grain size of 200 microns for the high speed rod rolling trials.

5. Low C steels have been investigated by the University of Oulu and are interesting because (a) they are insensitive to stress relaxation; (b) dynamic and static recovery are difficult to distinguish; (c) they exhibit considerable retardation

of softening when the strain rate is changed abruptly; (d) thermal history can exert a significant influence on the structural evolution e.g. the formation of homogenous polygonal ferrite on cooling at 5°C/s to 500°C followed by heating to 700°C.

6. The duplex stainless steel 2304 has been included in the multilevel model developed by CSM, having sharply contrasting properties to the medium C steel mentioned above, viz. zero recrystallisation and hence an accumulation of strain throughout rolling.

It has been supplied to CEIT for torsion testing and has been investigated in both as-cast and wrought conditions. These have significantly different structural properties. In the wrought state, the austenite and ferrite occur in incoherent parallel bands; during deformation the austenite tends to act as rigid fibres in the soft ferrite, thereby accommodating the torsional shear strain and producing linear strain hardening. In the as-cast condition, the austenite is mixed with, and coherently attached to the delta ferrite grains. During deformation, the austenite blocks tend to align with the direction of the deformation and to become separated from the ferrite giving rise to damage and induced softening.

Differences between the stress-strain curves for reversal are small except at high strains. The stress –strain curves are however significantly different from those observed in single phase materials. Moreover, the dissolution/rebuilding of the dislocation substructure during the transient is unobservable because of the different mechanical behaviour of the two phases.

7. Type 316 L stainless steel has been used by Corus in the experimental testing and modelling of multidirectional compression and H-V rolling. Optical metallography of specimens before thermomechanical processing has indicated the presence of delta ferrite with some twinning of the grains with evidence of inclusions and, in the interdendritic spacing, of the original as-cast microstructure. This causes difficulties in the initial definition of the grain size, ostensibly low at around 9 microns, and in its measurement at stages in its evolution. This was borne out in some of the optical metallographs in the H-V rolling trials, with a wide variation in grain size in the material being manifest even after a homogenising heat treatment prior to the trials.

Unlike the duplex stainless steel 2304, Type 316 stainless steel recrystallises, albeit moderately slowly. There is some evidence from the trials that the recrystallisation is retarded locally in the stock that has undergone an interpass turn but overall it was considered that, if occurring, this is a secondary effect. Indeed, for some of the tests, it was not possible, using optical metallography, to establish whether or not recrystallisation was substantial or non-existent in tests involving either REX promoting or REX inhibiting reheats. These findings were also reflected, because of the poor image quality, in IQ maps obtained by EBSD. Pole figure maps however give some indication of very slight recrystallisation.

8. A number of interstitial free steels have been examined by the University of Oulu, inc. Ti-IF, Nb –Ti IF. 'Strain-rate drop' tests on the Gleeble have shown that the final flow stress is higher than that at the lower strain rate but below that at the initial high strain rate. For Ti –IF the relaxation kinetics occurs over a longer time than for the austenite. Similar relaxation curves are found for the Nb–Ti-IF but the recrystallisation in the ferrite is not apparent. Examination of the dislocation

structure using electron microscopy, following the strain-rate drop tests, have shown that larger cells are formed than when the strain-rate is maintained at the constant higher rate, but that the misorientation of the grain boundaries is not affected. Analogous examination for strain reversal tests demonstrates also larger cells but a change viz. a decrease in the average misorientation of the grain boundaries.

9. Strain reversal on Armco iron specimens has also been studied by the University of Oulu and the tests showed it gives rise to considerably decreased rate of SRX and grain sizes much coarser than those for austenite.
10. CCrV steel is a tool steel, primarily of interest to the NORDIC community and the rolling of such steel has been modelled by MEFOS. For use in the model, flow stress data has been sampled by SIMR at strain rates between 0.8 and 30 s⁻¹ and at temperatures between 800 and 1200°C.

3.4 WP4 Formulation of Constitutive Laws

In the evaluation of constitutive equations (Task WP1) described in Section 3.1, it was established that the models generally employed were intrinsically limited in their analysis of the effects of complex loading paths, specifically not including the coupling between the local structural variables and the thermomechanical variables. It was also emphasised that reliable and accurate experimentation is vital in demonstrating the path dependence and for providing constitutive data for the models.

Many of the experimental results show a dependence of the flow stress behaviour on the strain path, in particular that the flow stress lags behind that predicted by an equation of state. There may also be changes induced by the deformation history in the grain size related to the recrystallisation event because the extent and speed of recrystallisation can also be affected by the deformation path. At a deeper level, other structural changes occur e.g. in dislocation density which influences recrystallisation rate or changes the hardness. To gain a more complete understanding of the forming process such structural variables therefore need to be included when formulating mathematical models for the process. Various types of models can be constructed, and many possible combinations of variables can be included as mentioned earlier. This variety is reflected in the models of the various partners.

1. It is possible, by a number of means, to elicit some of the effects of deformation history, bypassing the coupling between the thermomechanical and structural variables, one of these being the capture of experimental data on such effects and the input to a database and incorporation of relevant information into updated numerical models. An example of this approach is the use of visioplasticity in conjunction with correlation based image processing techniques to investigate the effects of reversal in plate rolling. These techniques enable, for various F-F & F-R schedules, the strain and stress fields to be calculated from examination of the deformation of a grid marked in the stock. This information has then been incorporated in FE models applied to such processes.
2. An empirical model of the complete stress- strain curve (including the transient) for strain reversal has been constructed from experimental assessment of several parameters. Some of the equations are sensitive to small changes in some of the coefficients but good agreement with experimental results is obtained provided that filtering techniques are applied to the data.

3. In order to better understand the underlying factors controlling the evolution of the flow stress, in particular, its lagging behind the mechanical equation of state following abrupt changes in strain rate, the University of Oulu has derived stress strain curves by adaptation of the analytical model developed by MEFOS relating the flow stress, under constant deformation conditions, as well as under decreasing strain rate to the total dislocation density. In adapting the model, several assumptions had to be made in order to establish the duration of the transient, the remobilisation curve for immobile dislocations and the evolution of the friction stress in this period etc. The stress strain curves could be fitted well with those measured. A model somewhat similar has been applied by the University of Oulu and CSM to investigate DRX and MDRX in rod rolling.

The models described above are useful for describing the stress behaviour even though they do not provide any direct information on the microstructural evolution associated with the complex strain path.

4. A sophisticated analytical model for the fraction recrystallised, including the effects of strain reversal, has been developed by CEIT using an Avrami equation in which the dependence of $t_{0.5}$ on strain path is taken into account. This is achieved by expressing the stored energy driving the recrystallisation (to which the above time is related) as a sum of contributions from the internal dislocation density and the misorientation across the cell boundaries; the balance between these contributions depends on the reversal and is determined from the pre-stress.

Numerical models predicting the distribution and evolution of dislocation density and other structural variables in conjunction with those of stress and strain etc have also been developed and are described below.

5. Corus has developed, in conjunction with Birmingham University, a unified viscoplastic model which takes into account the interaction between dislocation dynamics, grain size evolution, recovery, recrystallisation and isotropic but not kinematic hardening. The primary variables are thus the relative dislocation density, instantaneous grain size, the volume fraction recrystallised and the strain and stress tensors.
6. MEFOS has also formulated a model in conjunction with SIMR and MIKRAB based on dislocation density and vacancy concentration and including the effect of both these quantities on recovery and recrystallisation. A major softening process is recrystallisation; so that both metadynamic and dynamic recrystallisation have been incorporated by means of a sophisticated model developed in conjunction with SIMR and MIKRAB which also includes nucleation, growth and coarsening of the new grains. Notably, nucleation is not described using classical theory (which is inappropriate for recrystallisation) but is based on the abnormal growth of subgrains, embryos growing normally being unlikely to nucleate into mature grains. Grain growth at the metal surface and in the interior is considered separately, the driving forces being the reduction of stored energy due to reduction of grain boundary area, the latter parameter acting as a restraint at the surface. Coarsening is the mechanism for the evolution of the interior grains. After recrystallisation, the recrystallised and non-recrystallised parts behave differently concerning DRX. The model is also applicable in principle to multi-step deformation.

The flow stress is given in terms of the dislocation density by means of a modified Hirsch equation. The associated constitutive equations have been incorporated into an elastoplastic finite element model.

Although dislocation based models such as those above are potentially very useful in providing an understanding of material behaviour at microscopic and possibly sub-microscopic levels, they require many experimental data to realise this potential and provide an impetus for obtaining equipment and developing techniques for the acquisition of these data. Even without all these data they can nevertheless be used to carry out sensitivity and 'what-if' analyses.

Alternatively, a 'top-down' approach can be followed by extending a standard mesoscopic model to progressively deeper levels by augmenting the number of variables to include quantities related to the microstructural evolution such as fraction recrystallised, back stress and hardness,

CSM has formulated a constitutive model constructed at several levels incorporating respectively hardening and dynamic recovery, dynamic recrystallisation, and static softening, using experimental data obtained from torsion tests at CEIT and compression tests at the University of Oulu. At the first level, the model is a modification of the classical inverse sinh law in which the flow stress is expressed not only as a function of the primary variables - strain rate and temperature combined in the Zener-Hollomon parameter, but also of a strength or hardness factor, denoted by s where s is a hereditary variable, dependent on the strain history and evolving according to a prescribed differential equation. This factor corresponds to the hot strength reached without dynamic recrystallisation. The effect of the latter is incorporated at the next level via a secondary hereditary variable, viz. the DRX volume fraction evolving, when activated, according to an Avrami type equation dependent on strain, strain-rate and temperature. It is determined assuming that the structure responds instantaneously with the changing local conditions. The interpass (i.e. post deformation) softening is calculated, taking into account both static and metadynamic recrystallisation, using a model provided by the University of Oulu for axisymmetric compression. To apply to rolling schedules, procedures have been devised for identifying and computing prior strain rate measures to apply to the process. The effect of softening on flow stress is then calculated using the method of mixtures. The effect of strain path, in this case shear strain reversal, is introduced at a further level by locally changing the yield surface following such a reversal. Two hypotheses have been considered for defining this change viz. the flow stresses reduce uniformly (SSRS) or that only the shear stress component that underwent reversal is reduced. The model thus maintains the assumption of instantaneous response but takes into account the metallurgical evolution.

3.5 WP5 Incorporation of Constitutive Laws into FE Models of Rolling

Incorporation of a constitutive law into a finite element model of rolling requires the following steps:-

1. Choice of constitutive equation including combination of variables appropriate for the process considered, if several options are available.
2. Experimental processing of a sufficient number of specimens of the materials allocated over a range of conditions, related to the independent variables of the constitutive equations and representative of the rolling conditions considered, to

accurately determine the required constitutive response, i.e. in this case the flow stress.

3. Suitable software e.g. based on nonlinear regression or genetic algorithms to filter these data and to determine the coefficients of the constitutive equations.
4. Choice of suitable numerical techniques for the solution of these equations, in practice, efficient numerical techniques for the accurate integration of the rate equations.
5. Incorporation of these equations and the techniques for their solution into a form that is recognisable by the FE package employed i.e. corresponding to one of the listed types of deforming material or, more commonly, into a user subroutine in the format stipulated and written in an acceptable language.

The constitutive data have been incorporated in different ways by the respective partners reflecting the constitutive models adopted and the experimental facilities accessed.

1. In the load model used by TU Freiberg, the method of dimensions (i.e. the PI theorem) is used to relate the ratio of load to mean yield stress multiplied by the contact area to a minimal number of dimensionless variables aspect ratios of the rolling geometry and the coefficient of friction. The mean yield stress is determined from the stress strain curve (for the reduction, strain rate and temperature operative) by an appropriate integration procedure. The method of representing the stress-strain is left open. FE codes are also used as mentioned above into which constitutive data for a range of materials may be input together with information obtained by the viscoplasticity technique.
2. A vast amount of constitutive data for a range of materials deformed under several types of complex loading conditions has been generated using the Gleeble at the University of Oulu. These have generally been presented graphically but software based on regression analysis is available to fit equations if required. Constitutive models using the data for rolling applications in this project have been developed by other partners primarily CSM. However, models for ferritic rolling can be developed in future using the data obtained for materials such as Armco iron and low carbon steel.
3. Similarly many data have been obtained by CEIT for a range of steel qualities and used in the model developed by CSM. Moreover, the approach used by CEIT to construct equations of the flow stress curve can be usefully employed by partners in FE rolling applications.
4. Corus, for input into ABAQUS, have used, in the main, constitutive data in tabular format obtained from a variety of sources – results obtained on the local Gleeble, the literature, using the commercial software package ZMAT and more recently the local materials database. Alternatively, equations available in ABAQUS such as the overstress power law have sometimes been used. The dislocation based constitutive model formulated in this project, described in the previous section (and the viscoplasticity model formulated in the previous project and also used in this project) are incorporated into ABAQUS by either a UMAT or VUMAT routine. The constitutive equations in the previous model presented problems for some materials and applications. Firstly a good fit was obtained only for part of the range of strains etc and secondly the constitutive equations

are 'stiff', causing the standard backward difference methods employed in the numerical integration to converge extremely slowly or even diverge for some applications. In the recent model, flow stress data were obtained for Type 316 stainless steel from delayed double hit tests on the Gleeble. GA software has been used to fit the coefficients in the model equations for application to the H-V bar rolling trials and predictions are readily obtained using the computer hardware currently installed.

5. In the MEFOS modelling activity, CMnNb and CCrV steels have been considered. Flow stress data for the latter quality have been obtained from SIMR as described in Section 2.3. Flow stress data for CMnNb steel have been collected from the pilot plant at TU Freiberg at temperatures between 800°C and 1200°C and at strain rates of 0.1, 1 and 10 s⁻¹. These values are compared with corresponding values obtained from solutions of the constitutive equations obtained for a comprehensive set of combinations of values of the coefficients appearing in the constitutive equations. The coefficients chosen for the model are those providing the closest fit of theoretical and experimental data, in the topology defined by the root mean square error. Various optimisation algorithms for minimising this error using direct search, gradient methods, Newton-Raphson and quasi-Newton techniques have been explored as the ease with which particular functions neighbouring the data are found can be significantly dependent on the search technique employed (the latter method is generally efficient as it does not require updating the Hessian matrix). Difficulties have been experienced in getting the best fit, especially for the CMnNb steel but also for the CCrV steel, related to fitting polynomials to the frictional stress at very low strains for which data were not available. Some difficulties can be avoided by the use of screening to eliminate spurious data, interpolation to include 'missing' data regions, filtering to smooth oscillations (as observed for the strain rate) and normalising to handle wide variation in magnitudes of the respective gradients.

Problems were encountered in compiling and linking the SIMR and MIKRAB recrystallisation software with LS Dyna on the current pc platform, which were solved only through contact with the proprietor of the latter code, Livermore Software Technology Corporation.

The time integration of the constitutive equations is achieved using the Euler method since small time steps are used in the explicit LSDyna FE package installed on the computer platform recently installed; this gives results up to 100 times faster than the Runge-Kutta method used previously, without loss of accuracy.

In the plasticity model, maintaining points in stress space on (or below, if elastic) the stress surface has caused problems, as the radial return method previously employed is not implemented in LS-Dyna which, to be solved, have required direct introduction by the user of procedures such as penalty methods.

The constitutive model was incorporated into an FE code intended to model the forward-reverse rolling of plates at Freiberg. Consequently, it had to include features such as heat transfer viz. radiation, convection and conduction and interpass behaviour. Conduction and thermal contact have proved to be inefficient in searching for elements in contact and to give only small improvements to the solution and therefore are not currently incorporated. Modelling of the interpass behaviour has proved time consuming because of the

small time steps used in the explicit LS-Dyna code but, as remarked above, is very efficient in modelling the deformation.

6. The constitutive model developed at CSM is capable of operating at several levels, in particular with or without recrystallisation and shear strain reversal, it is capable of handling many of the structural changes occurring in a hot rolling schedule and can vary markedly depending on the type of rolling process and the steel quality rolled. In particular, it can model; (a) the rolling of bars of medium-C steel for which recrystallisation may be very fast resulting in annihilation of strain history and associated with grain refinement and there is also a reversal of shear strains which modifies the DRX (b) the rolling of stainless steel plates, at relative low temperatures at which the recrystallisation is non-existent so that normal strains accumulate throughout the passes and grains elongate. The shear strains may reverse but there will be some accumulation of residual shear strain.

Accordingly, it has been incorporated into ABAQUS FE coupled temperature displacement models of bar and flat rolling.

Constitutive data for the medium C steel have been obtained from results of torsion tests at CEIT and compression tests at the University of Oulu with comparable hardening rates.

- (a) Shear strain reversal in torsion produces material softening; this has been implemented as an isotropic contraction of the yield surface.
- (b) Following a change in strain rate, the steel almost follows an equation of state, with a strain transient of about 0.02. Analogous delay in the microstructural response may have a significant effect on dynamic recrystallisation especially in bar rolling where the interpass times are extremely short.

The data for the duplex stainless steel have been obtained from torsion tests at CEIT. Shear strain reversal has been modelled using both hypotheses.

For both steels shear strain reversal produces material softening and accordingly has been incorporated into both constitutive equations.

- (i) The model was used to assess the amount of normal strain reversal occurring between the H-8th and V-9th passes of the Ori-Martin bar mill in the rolling of the fast recrystallising 55Cr3 medium C spring steel. Reversal of the shear strain is predicted in the rolling plane within the 8th pass. Predictions using the selective softening hypothesis for reversal differ very little from the those obtained using the classical isotropic hardening law; alternatively, use of the full softening hypothesis has no effect on the rolling forces but a significant effect on torque. The shear strain reversal influences the process of dynamic recrystallisation especially towards the rolling stock surface where the DRX volume is reduced, which occurs in a central but wide area of the bar, therefore having minimal effect on the shear strain near the roll stock surface. The presence of such

an alternating H-V stand reverses the strain components in directions transverse to rolling, so cancelling accumulation effects. The only cumulative strain component is the one in the rolling direction; since recrystallisation is completed almost everywhere in the interpass period, the effect on the 9th pass is minimal and prevents any further reversal.

- (ii) Levels I and IV of the model have been applied to the flat rolling of the essentially non-recrystallising duplex stainless steel 2304 in roughing and tandem configurations. The shear strain therefore accumulates from pass to pass, especially for the reversing stands; the extent of the area affected is however almost the same. Use of the full strain hypothesis is found to be too excessive on softening and consequently has a significant effect on torque and power, predictions of roll force and power in each second pass being higher when the NO-SSRS hypothesis is employed. Predictions for the tandem mill are noticeably higher than those for the reversing stand but, when compared under the same thermal conditions, the difference, in contrast, becomes marginal. In contrast with results for the bar mill, these predictions differ significantly from those for the first pass. (Comparison with values measured in the mill is covered in Section WP6).

3.6 WP6 Model Validation

1. Finite element models developed by Corus have been applied to selected experimental tests and rolling trials for all the types of complex loading path considered. Where measurements have been possible, predictions have generally been in good agreement e.g. for loads and torques and spread in the forward-reverse plate rolling trials. Quantitative comparison of strains has not been possible because of unsuitability of the Hensler-Gifkins technique for the metallurgical structures employed but normal strains e.g. for the multidirectional compression tests are qualitatively as expected. Application of the dislocation based model to the bar rolling trials predicts: (a) for pass 1, an essentially uniform distribution of grain size and dislocation density, but not of Mises stress, this being maximum at the quarter depth in the mid-plane position; (b) little recrystallisation after the interpass as suggested by the experimental results.
2. The TU Freiberg model, including subroutines for the stress-strain behaviour and implemented using the FE code MARC has been applied to the plate rolling schedules considered for CMnNb steel. Differences in the shear deformation during the second pass have been indicated between forward and reverse rolling. The flow lines during the passes have been computed from the predicted fields and compared with results obtained using visio-plasticity measurements. The deformation of the flow lines and the wire grid are very similar apart from small deviations near the exit of the rolling gap in the middle of the rolling sample as well as in an area located at a quarter of the sample thickness below the surface. The deviations are probably due to the out - of - plane displacement of the specimen surface, frictional effects, the influence of thermal boundary conditions and the use of a 2D rather than a 3D FE model.

From computational studies the origin of the inhomogeneity and differences in recrystallisation behaviour have been identified.

3. The MEFOS model has been validated by applying it to compression and to rolling. The first application was set up to validate the flow stress behaviour i.e. to check that the data had been sampled correctly. FE predictions, at several temperatures and strain rates, of the development of the yield stress during the compression were compared with curves obtained from the constitutive model. Good agreement was obtained for CCrV steel, and for CMnNb steel except near the plateau and where the second derivatives are the largest i.e. the model does not explain the softening phenomena at strains around 0.4. The second application was to assess the model for the benchmark application i.e. the plate rolling trial at TU Freiberg. Predictions indicate; (a) a maximum effective stress within the deformation zone; (b) relaxation in the rolling direction; (c) an increase in dislocation density during rolling; (d) higher effective plastic strain at the middle than at the front section; (e) almost constant vacancy concentration.

High values of dislocation density are predicted by the model; it has not been possible to confirm these values experimentally.

4. The constitutive model developed by CSM has been set up using data for 55Cr3 steel from the Gleeble compression machine at the University of Oulu and from the torsion machine at CEIT and has also been validated experimentally. In the first case, for a test temperature of 1050°C and a strain rate of 1 s^{-1} excellent agreement was obtained between predicted and experimental loads and stress-strain curves. The torsion tests used the same primary variables as the Gleeble and comparison was made between corresponding values of torque. Very good agreement was obtained for strains relevant for rolling i.e. up to 0.7. Predictions of strain obtained from the FE model also agreed well with those calculated using the analytical Fields – Backoffen model. Stress-strain curves obtained using the FE model fit well with those obtained using the analytical model. Results were subsequently compared with test results on the Gleeble with abruptly changing strain rate conditions. Differences in stress were found, indicative of the errors made by assumption of instantaneous response in the model; these in general were small, results for the material occurring at a strain of c. 0.02 later than predicted. The model was then validated for rolling both on experimental and industrial mills. Reasonable agreement was obtained for the loads. For industrial validation, the ORI-MARTIN mill was chosen. Predictions of load evolution are in excellent agreement with the trace from the mill.

4. Conclusions and recommendations

The objectives of the project, in conformity with the Technical Annexe, have been achieved through the sharing of expertise, helpful co-operation of all the partners and synergies resulting therefrom. Detailed conclusions regarding partners' activities are given in Appendices 1 to 6. General conclusions regarding the tasks are given below.

4.1 Modelling

A review of constitutive models has shown that, generally, those previously employed are intrinsically unsuitable for providing a comprehensive understanding of the influence of complex loading paths since they do not include the coupling between local quantities

describing or closely affected by the structural behaviour and the thermomechanical variables. They may however be useful in providing a partial understanding of the effects of path e.g. on shape and in assisting with designing schedules etc.

Modelling the influence of deformation history on macroscopic quantities such as shape can be carried out, without including the structural quantities, either by skillfully incorporating information, gleaned from observation, into the input data of the model, updating the latter as necessary or using more complex forms of constitutive equation (e.g. of integral or functional type). The first approach has been adopted by TU Freiberg, by means of online viscoplasticity in conjunction with a correlation based image processing technique and has obtained close predictions of the effect of reversal in the plate rolling trials.

Modelling of the structural coupling can be achieved either by a 'bottom-up' approach in which the structural behaviour throughout the stock is described using a microstructural model, to which mesoscopic variables such as stress are related or, by a 'top-down' approach in which a multi-level model is progressively developed from a standard mesoscopic model by the successive augmentation of the set of variables to contain quantities closely related to the structural evolution. The first approach, in principle, provides a comprehensive description of the processes in the stock but requires extensive and expensive acquisition and processing of data. The second approach has a well defined but limited scope, can predict associated quantities relatively easily with a moderate amount of data and can be extended to other levels.

The first approach has been adopted by MEFOS in the development of a sophisticated fundamental model based on dislocation dynamics, vacancy generation and recrystallisation including nucleation based on abnormal grain growth (developed in conjunction with SIMR and MIKRAB) from which the flow stress is derived by a modified Hirsch equation and incorporated into an elastoplastic FE model. Many data for yield stress over a range of strains, strain-rates and temperatures, etc. have been collected for CMnNb from Freiberg and CCrV steel from SIMR. Problems were encountered in choosing appropriate coefficients of the constitutive model because of difficulties in fitting a non-analytical (i.e. integrated from a system of differential equations) function to experimental curves, possibly reflecting limitations in state of the art optimisation methods. Predictions of stress, strain etc. and the microstructural quantities have been obtained for a compression test (to validate the model) and also for the benchmark application of plate rolling at Freiberg. Reasonable predictions of the thermomechanical variables have been obtained but verification of structural quantities such as dislocation density was not possible.

A similar approach, but involving fewer variables, was followed by Corus in adopting the dislocation based model developed in conjunction with Birmingham University. Unlike previous models, it efficiently integrates the constitutive equations and predicts stress, strain, grain size evolution, fraction recrystallised and dislocation density. It has been applied to the H-V rolling trials of Type 316 Stainless steel using data obtained from delayed double hit tests on the Gleeble. It predicts very little recrystallisation on entry to pass two as indicated by the experimental rolling trials.

The second approach has been followed by CSM in the development of a model describing hardening, dynamic recovery, dynamic recrystallisation and static softening. A major aim of the model is to investigate differences between shear reversal and normal reversal in hot rolling. It has been applied to rolling of both a fast recrystallising medium carbon steel and a non-recrystallising duplex stainless steel using torsion and compression data from CEIT and the University of Oulu to determine the coefficients in the constitutive equations.

- The model has been validated using the above facilities even for conditions of changing strain rate.
- The model has also been validated on its local plate mill and on the ORI – MARTIN bar mill and gives rolling load and power predictions in agreement with measured values.
- In spite of differences in shear strain accumulation, minimal differences in roll force between roughing and tandem configurations are predicted for the non-recrystallising steel.
- For bar rolling of medium C steel, the model predicts DRX at the core unlike simplistic models, which predict SRX everywhere.
- It has not been possible to verify the structural predictions metallographically.
- Softening does occur in practice but not to the extent predicted by the model assuming full softening of the yield surface.

Useful non-numerical models describing the effect of strain have also been developed in the project.

- An empirical set of equations describing the entire flow stress curve for strain reversal has been established by CEIT.
- An analytical model of the flow stress curve following a change in strain-rate, based on solving a dislocation model of MEFOS with simplifying assumptions has been obtained by the University of Oulu and graphs obtained are in good agreement with its experimental results.
- An analytical model for the fraction recrystallised, including the effects of strain reversal has been derived, by CEIT, from the underlying theory based on dislocation density and grain boundary misorientation and gives approximately good predictions of softening except at the end of the plateau.

In application and further development of the models it is recommended that

- More comparison is made between predictions and measurements of microstructure and development of more accurate techniques for the latter is actively pursued.
- Techniques are developed for the efficient and accurate numerical integration of the constitutive equations.
- State of the art optimisation techniques are explored in order to determine coefficients which achieve a closer fit of the integrated solution of the constitutive equations to the experimental flow curve.

4.2 Experimental investigation and material characterisation

There is evidence from experimental trials that the complex loading paths considered, for certain values of the process parameters, had a significant influence on the flow stress behaviour and the microstructural evolution:

- In tension tests, an abrupt decrease in strain rate, in the dynamic recovery regime, gives rise to;
 - A lag in flow stress behind the behaviour of a mechanical equation of state.
 - An even greater lag in the static recrystallisation rate, especially after a strain resulting in recrystallisation (the values for the above close to, or higher than, those for the constant initial strain rate).
 - For ferrite, larger cells than those if the initial strain rate is maintained.
- In tension/compression and torsion, a strain reversal can give rise to;
 - A noticeable Bauschinger effect.
 - A transient characterised by a plateau.
 - A change in the SRX rate depending on the magnitude of the strains and for sufficiently large interpass times.
 - For ferrite, grain coarsening, larger cells and more low angle boundaries.
- In tension/compression and torsion, a strain reversal coupled with an abrupt drop in strain rate gives rise to;
 - A more pronounced Bauschinger effect.
 - A longer plateau.
- In torsion, a strain reversal coupled with an abrupt increase in strain rate gives rise to the opposite of the above effects for a decrease (Note: no relevant data available in this project for tension/compression).
- In tension/compression, re influence of temperature on austenitic steels;
 - Thermal history effects are insignificant (viz. a slight increase in flow stress for a rapid increase in temperature from 900°C to 1100°C).
 - Actual interpass temperature does affect recrystallisation.
- Multidirectional compression tests indicate that;
 - Multiple hits at constant temperature give slower early recrystallisation than a single hit of the same total magnitude and may inhibit REX entirely if, individually, are sufficiently small.

- An interpass turn of 90° hastens early recrystallisation when one of the reductions is large but significantly retards it when both hits are moderately large.

Even though results from experimental tests need to be qualified according to the parameters employed, significant effects of loading path have frequently been observed. In contrast, effects from rolling trials have generally been found to be secondary and in some cases, the evidence has been inconclusive or even conflicting.

- In forward –reverse rolling trials using CMnNb steel plates;
 - Noticeable differences between the deformation of stock rolled in the same direction and that for the stock reversed have been observed using visioplasticity techniques.
 - Positive and negative changes in load in 2nd pass for the two schedules have been observed depending on the temperature.
 - Differences in development of austenitic grain structure have generally been difficult to detect because of the presence of bainite etc, but, where martensite was produced, the evidence was that reverse rolling tended to reduce the grain size.
- In H-V rolling trials using Type 316 stainless steel bars;
 - The fraction recrystallised locally i.e. near the surface is smaller in stock that has undergone an interpass turn but the load following the interpass is smaller suggesting that recrystallisation is not widespread.
 - In trials investigating the effect of various interpass heating programmes with or without an interpass turn, optical metallography could produce no reliable evidence of significant or absent recrystallisation for both the interpass heating designed to promote SRX and that designed to inhibit it. Electron microscopy using EBSD produced IQ maps of poor quality and was therefore inconclusive. Pole figure maps, however, showed slight evidence of a little recrystallisation in both cases. OM also revealed wide variation in grain size and show that maximum grain sizes are greater in specimens that have been turned; the large differences however are more likely indicative of material inhomogeneity than retardation of grain refinement.
- In high speed rod rolling;
 - Slightly smaller grain sizes have been observed at the higher speeds and higher reheating temperatures.
 - More significantly, loads are higher at higher temperatures-probably because of phase transformation.

Inconclusive results from rolling trials could have arisen because;

- Effects were masked by other phenomena in particular related to heating and cooling.
- Difficulties in measuring the quantities required in relation to;
 - Measurement technology.
 - Material employed.
- Statistical variation.
- The deformation paths considered in a given trial are not sufficiently different from one another:
 - The strain rate changes after the roll bite are not as large as those considered in the tests.
 - Despite the apparent difference, the strain angle between the deformation paths is not large for the process considered – as remarked in the introduction.

To address these issues in particular, differences between results from experimental tests and rolling trials, the following recommendations are therefore made in implementation of suggestions from the project and in future research.

- Reheating and cooling should be carefully controlled and monitored by integration and automation of heating, rolling and quenching facilities comparable to the set up in Gleeble machines. Instrumented rolling trials should be carried out and specially constructed refractory units should be used where approximately isothermal conditions are required.
- Difficulties in measuring quantities should be reduced by;
 - Use of more sophisticated technology such as EBSD etc using field emission rather than tungsten filament equipment, SEM etc.
 - Use of materials e.g. Fe30% Ni, with analogous mechanical properties to those of practical interest but more amenable to observation but without obfuscating phases from transformation and significant initial heterogeneity.
- Statistical variation should be reduced by; (a) the use of more samples taken from more positions of the rolled stock, for metallographic examination; (b) the taking of a more statistically meaningful sample by repetition of trials and the consideration of other combinations of parameters selected using DOE techniques.
- More complex loading paths occur in practice globally and locally during rolling and should be considered, in particular
 - Greater strain rate variation e.g. effects of roll acceleration/deceleration and skidding etc.

- Greater changes in strain path both globally e.g. in edge rolling, 'cross rolling' operations for plates and locally in the rolling of beams near the web-flange interface and in sections such as piling near the lock. The latter can be studied using FE and then reproducing the deformation paths predicted, in samples using an ASP machine to examine the structural evolution.

It is believed that the choice of loading path is important e.g. as a means of refining the grain size by recrystallisation and thereby increasing the strength. This could be important for plates and beams as an alternative to heat treatment or to increasing the carbon content which is detrimental to weldability. The loading path may also be significant when used in conjunction with novel or composite rolling shapes such as may be relevant in downstream applications, and with other material qualities more susceptible to changes in process history.

The project suggests means for studying experimentally and theoretically the effects of loading path, a study which should continue, even if down a different avenue.

5. References

1. Ingham, P M: 'The Effect of Strain Reversal and Strain-Time Path on Constitutive Relationships for Metal Rolling/Forming Processes', Final Report ECSC Project, 7210.EC/811, 2000.
2. Davenport, S B, Higginson, R L and Sellars, C M: 'The effect of strain path on material behaviour during hot rolling of FCC metals', Phil. Trans. R-Soc, London A (1999), 357, pp1645-1661.
3. Angella, G: 'Strain path, flow stress and microstructure evolution of a austenitic stainless steel at high temperature', Ph.D. Thesis (University of Sheffield), 2002.
4. Wynne, B: Dept. of Engineering Materials, Univeristy of Sheffield (Private communication).
5. "Process Modelling", Ch. 8 in 'Advanced Physical Chemistry for Process Metallurgy', Ed. Maeda, M, Academic Press, 1997.
6. Weinan, E and Engquist, B: 'Multiscale Modelling and Computation', Vol. 50, No. 9.
7. Truesdell, C and Noll, W: 'The non-linear field theories of mechanics in 'Encyclopedia of Physics', ed. Flugge, S, 1965, Vol. 3/3, Berlin, Springer-Verlag.
8. Coleman, B D: 'Thrmodynamics of materials with memory' Arch. Rational, Mech. Anal, 1964, Vol. 17, pp1-46.
9. Malvern, L E: 'Introduction to the mechanics of a continuous medium', publ. Prentice Hall, Inc. 1969.
10. ECSC Research Contract 7210.EA/805: 'Three-Dimensional Model for Metal Flow under Triaxial Strain Conditions', British Steel Corporation, 1983.

11. Hohenemser, R and von Prager, W: *Z Angew. Math. Mech.*, 1932, Vol. 12, pp216-226.
12. Oldroyd, J G: 'A rational formulation of the equations of plastic flow for a Bingham solid' *Proc. Cambridge Phil Soc.*, 1947, Vol. 43, pp100-105.
13. Perzyna, P: 'On the constitutive equations for work hardening and rate sensitive plastic materials', *Bull. Acad. Pol. Sci. Ser. Sci. Technol.*, 1964, Vol 12, (4), pp199-206.
14. Rice, J R: 'Inelastic constitutive relations for solids: An internal-variable theory and its application for metal plasticity', *J. Mech. Phys. Solids*, 1971, Vol. 19, pp433-455.
15. Lematre, J and Chaboche, J L: '*Mecanique des materiaux solides*', Dunod, Bordes, Paris, English edition, CUP, Cambridge, UK, 1985.
16. Chan, D S, et al: 'A survey of unified constitutive theories', 2nd symp. NASA LEWIS, Non-Linear Constitutive Relat. High Temp. Appl., Cleveland Ohio, NASA Conf. Publ. 2365, 1969.
17. Miller, A H, ed.: 'Unified constitutive equation for plastic deformation and creep of engineering alloys', Elsevier Applied Science, NY, 1987.
18. Krausz, A S and Krausz, K (eds.): 'Unified Constitutive Laws of Plastic Deformation', Academic Press, 1996.
19. Bodner, S R and Partom, Y: 'Constitutive equations for elastic viscoplastic strain-hardening materials', *J. Appl. Mech.*, 42, pp385-389.
20. Chaboche, J L: 'Viscoplastic constitutive equations for the description of cyclic and anisotropic behaviour of metals', *Bull. Acad. Pol. Sci., Sem., Sci, Ser.*
21. Dzubinsky, M: 'Comparison of recrystallisation kinetics established by stress relaxation, double hit, optical metallography methods and using orientation maps obtained by SEM/EBD', Report Code, STC/SMT MMM/KR/08616/2002/A, (Reference Source no. 107498), Corus Research, Development & Technology, Swinden Technology Centre, 2002.
22. Sellars, C M In: Sellars, C M, Davies, C J eds. *Proc. Conf. on Hot Working and Forming Processes*, London, Metal Society, 1980, pp3-15.
23. Karjalainen, L P and Perttula, J: *ISIJ*, 1996, Vol. 36 (6), pp729-736.
24. Karjalainen, L P: *Mater. Sci. Technol.* 1995, Vol. 11, pp557-565.
25. Hensler, J W and Gifkins, R C: 'The estimation of slip strain during creep', *J-Inst. Metals*, 1963-64, Vol. 92, p340.
26. Visual Mind 5.01 (demo version), Norcan, copyright 1998-2003.

Table 1: Programme bar chart

No.	Title	Category	Year 1 (1970-1971)		Year 2 (1972-1973)		Year 3 (1974-1975)		Total 1970-1975
			Number	Amount	Number	Amount	Number	Amount	
21	Expenditure of Central Government	General							
22	Expenditure of State Government	General							
23	Expenditure of Union Government	General							
24	Expenditure of Local Government	General							
25	Expenditure of Central Government	General							
26	Expenditure of State Government	General							
27	Expenditure of Union Government	General							
28	Expenditure of Local Government	General							
29	Expenditure of Central Government	General							
30	Expenditure of State Government	General							
31	Expenditure of Union Government	General							
32	Expenditure of Local Government	General							
33	Expenditure of Central Government	General							
34	Expenditure of State Government	General							
35	Expenditure of Union Government	General							
36	Expenditure of Local Government	General							
37	Expenditure of Central Government	General							
38	Expenditure of State Government	General							
39	Expenditure of Union Government	General							
40	Expenditure of Local Government	General							

Table 2: Matrix of interactions between partners

		Receivers					
Providers	Conus	Conus	CEIT	CSM	MeFos	TU Freiberg	University of Oulu
			CMnNb steel <i>Exptl observations re path change on REX Dislocation Model</i> Model of REX	Model of REX	Dislocation Model	CMnNb steel <i>Exptl observations re path change on REX Model of REX</i>	CMnNb steel <i>Observations re path change on REX Model of dislocation & REX</i>
	CEIT	Torsion Data for CMnNb steel FS Curve Model (revis) Observation re revis on REX/FS Dislocation Model		Torsion Data for Medium C & Duplex Steels Model verification FS Curve Model	Dislocation Model FS Curve Model	FS Curve Model	FS Curve Model Observation re revis on REX/FS
	CSM	Shear Reversal Model/ Def of σ_c strain & strain-rate REX Model	Medium C steel Duplex stainless steel Predictions for Torsion of Med. C steel		REX Model	Shear Reversal Model REX Model	Medium C steel Predictions for Compression of Med. C steel
	MeFos	Predictions of Rolling of CMnNb steel for Benchmark case	Model of Dislocation & REX	Model of REX		Predictions of Rolling of CMnNb steel	Model of Dislocation & REX
	TU Freiberg	Experimental Rolling of Rods (Fast & Slow) & Plates of CMnNb steel Experimental Rolling of plates of CMnNb steel +Met/Emprical/FE Analysis.	Torsion results for CMnNb steel	Exp/FE Shear Reversal Results	FS Data of CMnNb steel		Compression results for CMnNb steel
	University of Oulu	Compression Data for CMnNb steel Observation re revis on REX/FS	Compression results for CMnNb steel Observation re revis on REX/FS	Compression Data for Medium C steel Model Verification	Exptl results on dislocation structure	Compression results for CMnNb steel	

(Cells show resources provided or common interest {shown in italics})

Table 3: Programme matrix

	PARTNERS					
TASKS	Corus	CEIT	CSM	MEFOS	Freiberg	OULU
Fast $\Delta\dot{\epsilon}, \Delta\theta$						✓
ϵ reversal (Tension/Compression)	✓					✓
ϵ reversal (Torsion)		✓				
Bidirectional ϵ tests	✓					
ϵ reversal, $\Delta\dot{\epsilon}$ (Ferritic regime)						✓
H-V Rolling	✓					
Forward, Reverse Rolling					✓	
High Speed Rolling					✓	✓
Structures/Metallography	✓	✓				✓
Evaluation/Formulation of Constitutive Models	✓	✓	✓	✓	✓	✓
Constit. Models into FE Models of Rolling	✓		✓	✓	✓	
Validation of Models	✓		✓	✓	✓	

Table 4(a-f): Activities of partners

(a) Corus activities

Steel Quality	Type 316 stainless steel	CMnNb Steel
Loading Path	H-H/H-V (2 passes w/wo 90 degree turn)	Forward/Reverse Rolling (2 passes)
Workpiece	23 mm Square Bar	Plate
Temperature (°C)	~ 1100	1020 1050
Strain per pass	0.24-0.25	0.15 0.17 0.21
Mean strain rate (s ⁻¹)	5.4 – 5.9	8.7 9.6 10.1
Inter pass time (s)	10 - 40	10 -20
Models	REX/Grain Growth – empirical equations (see Ref [A1.5]) Unified Constitutive model inc. Hardening, Softening, Dislocation Dynamics (see Ref [A1.6])	
LOADING PATH	Multi direction Compression(in Gleeble) (1,2 or 3 passes)	
Workpiece	10 mm cube specimen	
Temperature (°C)	1000,1100	
Strain per pass (%)	0.25 –0.77	
Mean strain rate (s ⁻¹)	1	
Models	As for rolling	
LOADING PATH		Tension/Compression(in Gleeble)
Temperature (°C)		1000 1050 1100
Strain per pass (%)		0.2/0.2 & 0.4C
Mean strain rate (s ⁻¹)		1
Models		

(b) CEIT activities

Torsion	CMnNb steel	Medium – C steel	Duplex stainless steel	Duplex stainless steel
Condition	Wrought	Wrought	As-cast	Wrought
Temperature (°C)	850 - 1200	950, 1000, 1050	950,1000	950,1000.1050
Pre-Strain range	0.03-0.65 (up to 2 for monotonic)	0.1 and 0.2 (up to 10 for monotonic)	0.06,0.56,0.2	0.06-0.87
Reverse Strain	0.06-2	0.1-0.9	0.06 - fracture	0.06-0.87/fracture
Av, strain rate (s ⁻¹)	0.01,0.1,1.0	0.1, 1.0, 10	0.1,1.0	0.1,1.0
Multipass torsion sequences (>3 passes)	No	Yes, including reversion/or holding time between passes	Yes, including reversion between consecutive passes	Yes, including reversion between consecutive passes
Models				
austenite	Flow curves:Laasraoui & Jonas Phenomenological (dislocation dynamics):Bergstrom, Gil-Sevillano	FS by CSM SRV by CEIT - CSM	See Refs [A2.1-A2.7]	See Refs [A2.1-A2.7]
Ferrite				
Austenite+ferrite			FS by CSM-CEIT SRV by CEIT - CSM	FS by CSM-CEIT SRV by CEIT - CSM

(c) CSM activities

Rolling	Medium – C steel	Duplex stainless steel
Product	bar	strip
Temperature	1100°C- 850°C	1150°C – 950°C
Strain range	0.4- 0.8	0.18- 0.35
Mean strain rate	6 - 50 s ⁻¹	3 - 10 s ⁻¹
Models		
austenite	FS=CSM SRX/DRX=CSM-THE UNIVERSITY OF OULU SRV= CEIT - CSM	
Ferrite		
Austenite+ferrite		FS=CSM-CEIT SRV= CEIT - CSM

(d) MEFOS activities

Rolling	Medium – C steel	Duplex stainless steel
Product	bar	strip
Temperature	1100°C- 850°C	1150°C – 950°C
Strain range	0.4- 0.8	0.18- 0.35
Mean strain rate	6 - 50 s ⁻¹	3 - 10 s ⁻¹
Models		
austenite	FS=CSM SRX/DRX=CSM-THE UNIVERSITY OF OULU SRV= CEIT - CSM	
Ferrite		
Austenite+ferrite		FS=CSM-CEIT SRV= CEIT - CSM

(e) TU Freiberg activities

Steel Quality	CMnNb steel (supplied by Corus)
LOADING PATH	Rolling (2 passes Forward-Forward & Forward –Reverse)
Workpiece	Plate
Temperature (°C)	850, 900, 1000
Strain per pass	0.2- 0.7
Mean strain rate (s ⁻¹)	12 - 19 s ⁻¹
Inter pass time (s)	
Models	Empirical models for Flow Stress and Load, FE model
Experimental Techniques	Metallographic Analysis, Visioplaticity
LOADING PATH	Rolling (Up to 4 passes, Slow/Fast)
Workpiece	Rod
Speed (m/s)	15/30
Temperature (°C)	850 1100
Strain per pass (%)	0.5
Mean strain rate	60-140/370-740*
Models	Empirical models for Flow Stress and Load

(f) University of Oulu activities

Material			Armco Iron		Low C	
Loading path			$\Delta T, \Delta\epsilon'$	SRV	$\Delta T, \Delta\epsilon'$	SRV
Physical process:	Axisymmetric compression on Gleeble, tension					
Workpiece:	Cylinders/rods					
Met processes						
			FS	SRX/DRX	FS	SRX/DRX
	Metallography, stress relaxation					
Experimental Techniques	Optical microscopy; TEM, SEM-EBSD					
	Examination of grain/dislocation structures					
T (°C) austenite or ferrite (as appropriate)				800	700	
Strain-range				0.22/ 0.02-0.06	0.28/0.02	
					0.58/0.02	
Str rate (1/s)				0.1	1/0.01	
Models						
	Regression anal. of FS. Ferritic constitutive for rolling					

Medium C			IF Ti	
$\Delta T_{\Delta \epsilon'}$	SRV		$\Delta T_{\Delta \epsilon'}$	SRV
FS	SRX/DRX	FS	SRX/DRX	FS
	1000-1100	1050		
	0.2	0.2/0.02-0.6		
	0.1	0.1		
		800	800	
			0.3/0.02	
			1/0.01	

[illegible]

$\Delta T, \Delta \varepsilon'$ = Abrupt changes in strain-rate, temperature	SRV = strain reversal	FS = flow stress, SRX/DRX = static/dynamic recrystallisation

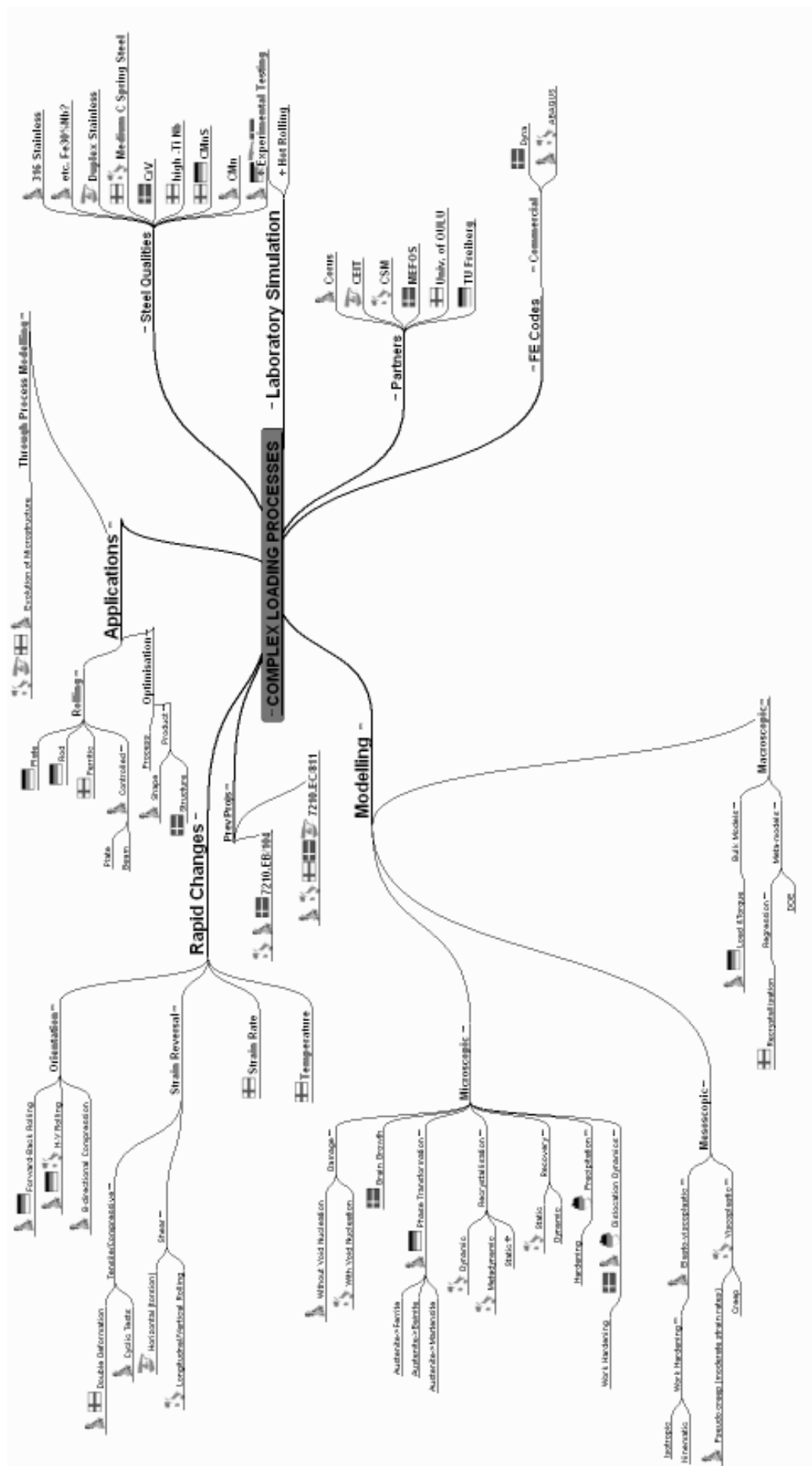


Fig. 1: Graphical conspectus of project constructed using mind mapping software [3]

Appendix 1

Detailed technical report by Corus UK Limited Authors: G J Watts, D C J Farrugia, B Cheong, Z Husain, and M Zhou

A1.1 Introduction

A1.1.1 Task allocation

Corus, in common with all the other partners is involved in reviewing constitutive models for the adoption or formulation of laws likely to be relevant for the materials and associated forming processes within their remit, particularly those involving complex loading paths viz. hot rolling of bar, plate and rod. In addition Corus employs at Swinden Technology Centre various experimental processes, as a means of readily providing data either for input to the models or for use in their validation. This equipment includes a Gleeble machine for testing in tension and compression, and laboratory bar and plate mills. Accordingly Corus has been engaged in the following work packages from the list given in the first part of this report and extracted from the Technical Annex.

WP1: Evaluation of Constitutive Models

WP2.2: Strain Reversal in Tension/Compression

WP2.4: Multi-directional Strain Tests

WP2.6: Stock Rotation during Rolling

WP2.7: High Speed Rolling

WP3: Material Characterisation

WP4: Formulation of Constitutive Laws

WP5: Incorporation of Constitutive Laws into FE Models

WP6: Model Validation

The contributions to these tasks are presented and discussed in Sections A1.2 to A1.7.

A1.2 Assessment of constitutive models (Task WP1)

Models of all the types mentioned in the main section of the report have been used by Corus in the simulation of hot rolling and, in particular, in the analysis of the stock constitutive behaviour and its interaction with the process parameters. The aim of this section is to provide details of the models considered towards fulfilment of the aims of the project. This involves both usefulness of existing models in the various tasks of the project and scope for development in analysing the influence of the strain path.

A1.2.1 Analytical and empirical models

These are frequently used to relate bulk or external quantities, such as rolling load and torque, and deformed geometry to the process parameters. Notable models adopted are those of Orowan [A1.1], and Ekelund [A1.2] respectively for prediction of rolling load and torque, and stock width on exit from the pass. Each of these models, being quite sophisticated, is applied via a pc program and instantly gives reasonably accurate predictions which can be used in design of rolling schedules. Orowan's program requires flow stress data for the steels under consideration; these may be experimental data or obtained from an empirical formula in terms of strain, temperature and strain-rate. It is often assumed that such equations derived for bulk conditions apply locally and can be used in FE codes for prediction of internal stress and strain distributions. Such use is questionable, as (a) structural factors such as grain size are not generally included explicitly and (b) it is assumed that the flow stress satisfies a mechanical equation of state i.e. that the flow stress responds instantly to deformation. Investigation of such formulae therefore forms a very important part of the project.

It may transpire, however, that use of such analytical flow stress models may give sufficiently accurate predictions under well defined conditions, not involving complex changes in strain path. Moreover improvements in accuracy can sometimes be obtained by adjustment of the formulae following comparison of predictions with experimental results.

Several analytical flow stress models have been used at STC, including the overstress power law and the Johnson Cook exponential law. Both are available in ABAQUS [A1.3] and are suitable for materials undergoing small deformation (i.e. hardening dominant) and subjected to small variations in strain rate and temperature.

The overstress power law model relates the stress and strain-rate viz.

$$\dot{\varepsilon} = D^* (\sigma^*)^p \quad \dots (A1.1)$$

where σ^* denotes the overstress i.e. the difference between the stress σ and the static or back stress σ^0 , ($\sigma \geq \sigma^0$) [A1.4].

It is assumed that the shapes of all stress-strain curves are similar to the hardening curve defined by the input table. This assumption is clearly not suitable for materials which undergo large deformation and exhibit softening due to recovery and dynamic recrystallisation over a wide range of strain rate and temperature. Moreover, the power law relationships between strain rate and the flow stress may not apply for some loading conditions where high stress is involved [A1.5].

In the basic Johnson-Cook analytical model, hardening occurs isotropically, the static yield stress is expressed by a power law, equation viz:

$$\sigma^0 = (A + B\bar{\varepsilon}^n)(1 - \theta^{*m}) \quad \dots (A1.2)$$

where $\bar{\varepsilon}$ is the equivalent plastic strain and θ^* is a dimensionless temperature parameter taking values between 0 and 1 (see [A1.3])

and the strain-rate dependence of flow stress is given by an exponential law.

Although the model uses a power function to represent the hardening curve, it cannot describe material behaviour exhibiting saturated flow owing to recovery or flow stress drop associated with dynamic recrystallisation at high temperatures. Similar to the power law, the basic Johnson-Cook model assumes that the stress-strain curves have similarity in shapes for different strain rate, which is not true for most experimental data.

The Johnson-Cook model has been extended to include strain-rate the latter varying exponentially with yield stress [A1.3] so that the latter is given by the equation:-

$$\bar{\sigma} = \sigma^0 \left(1 + C \ln\left(\frac{\dot{\epsilon}}{\dot{\epsilon}_0}\right)\right) \quad \dots (A1.3)$$

where σ^0 is given by Equation (A1.2).

The above Johnson-Cook model and overstress power law are suitable for high and low strain rates respectively. An alternative model shown to be satisfactory over a wide range of strain rates and temperatures [A1.6] is the Garofalo model [A1.7] in which the steady state strain rate is related to the saturated flow stress by the sinh law:

$$\dot{\epsilon}_{ss} = A [\sinh(\alpha \sigma_{ss})]^n \exp(-Q/RT) \quad \dots (A1.4)$$

This law is suitable especially for materials showing a change in predominant deformation mechanisms from low to high stress levels at elevated temperatures. At low stress levels and strain rates, typical of diffusion processes, the relationship is essentially linear. At high stresses and strain rates, the relationship is highly non-linear (approximately exponential), characteristic of processes controlled by dislocation mechanisms and so potentially useful in models of hot rolling. However, it does not model isotropic and kinematic hardening and recrystallisation [A1.8].

To investigate the effects of loading path on the initiation of recrystallisation another analytical flow stress model [A1.9,A1.10] and an alternative definitions of the equivalent strain have therefore been explored by Corus in the previous ECSC project [A1.8] in which a modified sinh law is adopted given by:

$$\dot{\epsilon}_{vp} = A [\sinh((\alpha(J_2(\sigma - x) - R_* - \sigma_0))^{np} (\sigma - \sigma_0)) \quad \dots (A1.5)$$

where $R_* = R_h$ (isotropic hardening) prior to dynamic rex and $R_h + R_s$ subsequently and x is a back stress associated with kinematic hardening.

Temperature dependence is not shown explicitly in most of the above equations but is implied, empirical equations being established for the coefficients, e.g. A , satisfying an Arrhenius equation. Also not included is the dependence on grain size e.g. via a Hall –Petch type equation.

Moreover, to predict the effect of deformation on the evolution of metallurgical structure, equations need to be incorporated for the initiation of development of recrystallisation and of grain growth. Several such formulae have been fitted empirically or derived theoretically based on the assumed grain geometry etc and stated in a previous ECSC project [A1.11], both for Type 316 Stainless steel and CMn steel.

For Type 316 stainless steel, the following formulae obtained respectively by Barbosa [A1.12] and Jaiswal [A1.13] for grain growth (microns) in terms of the time t (min) and absolute temperature T have been used in the current project.

$$\begin{aligned} d^2 &= [d_0^2 + 60t \times 3E13 \times \exp(-320000/(RT))] \\ d^6 &= [d_0^6 + 60t \times 1.78E8 \times \exp(-186698/(RT))] \end{aligned} \quad \dots (A1.6)$$

Similar formulae have been obtained for CMn steel by Sellars & Whiteman, and Leduc [A1.11]

For recrystallisation kinetics, a model has been used in which dynamic/static recrystallisation occurs according to the attainment or otherwise of a critical strain and evolves according to a JMAK law [A1.14,A1.15]. Generic formulae for the critical strain ε_c and evolution including the time $t_{0.5}$ for 50% recrystallisation taken from Ref. [A1.11] are reproduced in the equations below.

$$X = 1 - \exp(-0.693(t/t_{0.5})^k) \text{ when } \varepsilon < \varepsilon_c \quad \dots (A1.7)$$

where

$$t_{0.5} = A.Z^{-1}.d_0^m.\varepsilon^{-n}.\exp\left(\frac{Q_{\text{rex}}}{RT}\right) \quad \dots (A1.8)$$

and

$$\varepsilon_c = A_c.d_0^{l_c}.Z^{m_c} \quad \dots (A1.9)$$

with

$$Z = \varepsilon.\exp\left(\frac{Q_{\text{def}}}{RT}\right) \quad \dots (A1.10)$$

The static recrystallised grain size, d_{rex} is given by

$$d_{\text{rex}} = A_r.d_0^{m_r}.\varepsilon^{-n_r}.Z^{-1} \quad \dots (A1.11)$$

Specific values of the constants for Type 316 Stainless steel and CMnNb steel are given in Table A1.1(a-c). Other equations for Type 316 Stainless steel and CMn steel relating to recrystallisation and grain growth and also recovery may be found in [A1.11].

Temperature Equivalent Time for Recrystallisation under non-isothermal conditions

In obtaining the JMAK Equation (A1.7) for Type 316 stainless steel for the fraction recrystallised the Arrhenius rate term $z = \exp(-Q_{\text{rex}}/RT)$ is integrated with respect to time. Under isothermal conditions this integral is trivial viz. multiplication by the time elapsed. Under non-isothermal conditions however, the time dependence of the temperature T should strictly be taken into account. Consequently the above term is multiplied by an average i.e. temperature equivalent time t_{equiv} given by:

$$t_{\text{equiv}} = \int_0^t \frac{\exp(-Q/RT)}{\exp(-Q/RT_0)} dt' \quad \dots (A1.12)$$

where T_0 is the initial temperature for cooling (the final temperature for reheating) and T is dependent on the time t' . (effectively t_{equiv} is the length of the base of the rectangle corresponding to isothermal temperature T_0 , which is equal in area to the area beneath the Z vs time curve when the temperature varies.

To estimate t_{equiv} , a linear time dependence is assumed for the temperature i.e. $T = T_0 - rt'$

where $r = (T_1 - T_0)/t$ and T_1 in this case is the temperature at the end of the time interval.

Consequently

$$t_{\text{equiv}} = \exp(-Q/RT_0) \cdot \frac{Q}{RT_0} \cdot [\Gamma(-1, Q/RT_0) - \Gamma(-1, Q/RT_1)] \quad \dots (A1.13)$$

Using polynomial approximations in [A1.16] it has been shown that this expression can be approximated very closely by the simpler expression

$$t_{\text{equiv}} \approx T_0 / r \cdot [g(Q/RT_0) - g(Q/RT_1) \cdot (T_1 / T_0) \cdot \exp(-Qrt/RT_0 T_1)] \quad \dots (A1.14)$$

where g is the function defined by

$$g(x) = \frac{x + 3}{(x + 4)(x + 1)} \quad \dots (A1.15)$$

This formula has been incorporated into a Mathcad file for ready implementation.

A1.2.2 Numerical models

These are usually employed at the mesoscopic scale e.g. to determine internal distributions of stress, strain and temperature etc. A range of mesoscopic constitutive models are used at Corus and are implementable using the FE package ABAQUS, which is the principal commercial FE software package used at Corus to simulate the hot rolling process and is available in either ABAQUS/Standard and ABAQUS/Explicit [A1.17].

Three main classical plastic flow options within ABAQUS have been used by Corus viz. (a) elasto – plastic (b) viscoplastic and (c) creep with overlap between these options being possible in certain cases.

In elasto–plastic deformation, non-permanent deformation occurs prior to yielding. Although this is not usually significant in hot forming, elasto–plastic models of the latter have been developed. Such models developed at Corus assume the elastic deformation to be isotropic and so expressible only in terms of a Young's modulus and Poisson's ratio. Moreover, the metal is not usually taken to be perfectly plastic but to harden progressively with increasing strain. Generally, this hardening is assumed to be isotropic i.e. the increase in the yield stress is dependent on the magnitude of the strain and not its direction.

The default tabular input in ABAQUS [A1.17] has generally been used here because of its simplicity. The stress-strain data at various strain rates and temperatures are entered

following standard laboratory tests such as performed on a Gleeble thermomechanical simulator either in plane strain or in uniaxial compression mode. The flow stress at a given strain and strain rate is then determined from interpolation within the extreme boundaries of the data. No extrapolation is made for any strain, strain rate or temperature outside the defined conditions. However, during the rolling process, the strain rate varies continuously in the roll bite. Depending on the processing conditions (intermediate, finishing mills), it could be of the order of 10^2 s^{-1} or even higher, well above the strain rate conditions tested in most laboratory simulators such as on a Gleeble thermomechanical simulator, where the maximum achievable strain rate is approximately 25 s^{-1} . This could lead to underestimation in the calculation of the flow stress at higher strain rates during rolling. Consequently, the rolling loads and torques could be affected, although the effects on deformation such as spread and shape may be small [A1.18]. The influence on rolling load and torque can be a significant drawback when using these techniques for 'what-if' applications and roll design optimisation.

Alternatively, empirical equations defining a functional dependence of yield stress on the field variables can be incorporated into ABAQUS. For the Johnson Cook law (Equation (A1.3)), there is a specific option allowing this in ABAQUS. However, for isotropic hardening, such as described by the exponential law, the user subroutine UHARD must be used [A1.17]. ABAQUS also allows kinematic hardening in which the yield surface is translated or deformed according to both the magnitude and direction of the strain increment. For the metal forming processes considered in the project, kinematic hardening is not believed to be very relevant unless there is excessive strain reversal. It is important however for strain paths involving repeated cyclic loading as in fatigue testing. A kinematic hardening model [A1.19] has been developed at STC in ABAQUS/Explicit using a VUMAT routine [A1.8]; other such models are being developed in conjunction with Birmingham University [A1.20].

In viscoplastic flow, the material deformation is dependent on the strain rate. This dependence can be implemented by selecting the RATE-DEPENDENT option in ABAQUS and by including, in the deck for each selected strain rate value, tabular data relating stresses to strains etc. The model can then be integrated through time to give the deformation path and the stress-strain history. Other means are also possible of describing viscoplastic flow e.g. by specifying the ratio of the equivalent stress to the initial stress at various values of the field variables or by means of an equation incorporated using the user subroutine UHARD. Various viscoplasticity constitutive models [A1.18] have been developed at Corus representing the flow of steels at high working temperatures. The temperatures should include the adiabatic contribution from mechanical working and thus represent the flow softening associated with dynamic recrystallisation. This model has been validated for a concast steel Grade 43 and satisfactorily represents the work softening caused by dynamic recrystallisation.

Creep deformation is another means of simulating viscoplastic flow, one of whose key equations, viz. the overstress power law, Equation (A1.1), is formally similar to that for viscoplastic flow but with significantly different coefficients. This law can thus be included by invoking the CREEP option but with coefficients relevant to viscoplastic flow.

All the above models, however, do not explicitly take into account the microstructure. Although microstructural dependence of flow stress etc can be introduced via the coefficients and the local effect of stress, strain and temperature on fraction recrystallised etc. can be determined using equations such as those in Section A1.1 implemented in ABAQUS via a UVARM or the calculator utility, such approaches are likely to give a limited or even inaccurate picture of the influence of loading path on microstructure as they assume

the behaviour at the microscopic level is the same as that observed at the macroscopic level and, more significantly, neglect its interaction with that at the mesoscopic level. Coupling of the mesoscopic and microscopic behaviour requires more sophisticated approaches as discussed below.

A1.2.3 Augmented mesoscopic models

More complex constitutive models, with additional 'mesoscopic' variables implicitly containing information of the microscopic behaviour and satisfying a variety of functional relationships can be introduced [A1.9,A1.21]. Examples of such variables are (a) Critical strain for dynamic recrystallisation – initiated as described above (b) Drag stress -softening associated with dynamic recrystallisation (c) Back stress X_{ij} . kinematic hardening described by the developed by dislocation pile-up and retarded by dynamic recovery (d) Isotropic hardening. These satisfy a hyperbolic flow rule typical of dislocation controlled deformation involving the back stress and isotropic hardening and flow softening caused by dynamic recrystallisation. Such models can be constructed in ABAQUS/Standard using the complex user subroutine UMAT [A1.8]. Indeed, the above additional variables were incorporated into a thermo - elasto - viscoplastic model by Dunne [A1.10], adopted in the previous project [A1.8]. The equations for this model are given in the final report of the latter. With variables (b-d) set to zero and the critical strain effectively set out of range of the strain predictions, the traditional elasto-plastic model is obtained. It was decided, to explore the viscoplastic constitutive model described above both with and without the options of isotropic and kinematic hardening and softening.

The determination of the coefficients appearing in such models, however, is itself a significant issue in the formulation of constitutive models; It can be achieved in a number of ways. (i) They can be estimated empirically by obtaining a closely fitting equation (using regression or genetic algorithms (GA)) to load data determined experimentally at various strain-rates etc. This approach however suffers from the drawback that the data relate to bulk rather than mesoscopic deformation. (ii) Alternatively, they can be estimated by optimising the fit of FE predictions of these parameters obtained using various sets of the parameters. Although this may generate a model giving accurate predictions of bulk parameters it may not be suitable for predicting the internal stress and strain distributions and structural properties such as grain size which are partially dependent on the latter.

Augmented mesoscopic models such as those above cannot completely represent material behaviour since they do not explicitly contain microscopic variables and so cannot their evolution and interaction. To understand these processes fully, it is necessary to relate directly the mesoscopic variables to the underlying mechanisms of dislocation dynamics and to incorporate these relationships into a phenomenological model [A1.22] containing the relevant internal variables and additional equations.

A1.2.4 Phenomenological Models

A significant internal variable in these models is the dislocation density. Dislocation dynamics models use dislocation density as a history dependent internal variable [A1.23,A1.24] to represent the microstructural changes and macroscopic response of flow stress during hot rolling. They take account of the generation, accumulation and interaction of dislocations (both mobile and immobile) and their annihilation through recovery and recrystallisation. Combined use of dislocation density based and viscoplastic models provides a good means for handling complex loading, since they can predict complex history-dependent and time-dependent material behaviour including changes in flow stress and are able to trace the strain path.

Corus has, accordingly, utilised the dislocation model developed by MEFOS, which takes account also of recrystallisation and the vacancy concentration [A1.8] incorporating it into an ABAQUS FE program for hot rolling of steel by means of a UMAT user subroutine. Problems of numerical stability however, arise when integrating the equations unless extremely small time steps are taken. The method of implementation within ABAQUS v6.3 is also being assessed viz. using the *CREEP card or using the UMAT routines, the former option being the easier. For the latter, studies have been commenced into incorporating the exact Jacobian instead of the elastic D matrix used previously. It is considered that this allows larger, reasonable time steps to be used with good accuracy.

Another key variable is the grain size. The unified constitutive model has been developed by Birmingham University [A1.20] in conjunction with STC. Based on physical mechanisms for evolution of grain size in addition to dislocation dynamics, recovery, dynamic and static recrystallisation and their coupling with viscoplastic metal flow, this model has been explored in the latter stages of the project. It is frequently assumed that the metal prior to rolling has a uniform grain size, the stock having been adequately soaked in the reheating furnace. In practice, this is found not to be the case, metallographic examination revealing grains of various sizes distributed throughout the stock and even locally a statistical distribution occurring of significantly large standard deviation. It is generally assumed that this distribution is lognormal [A1.25]. If the mean and standard deviation of this distribution are denoted by $\bar{d}$ and s respectively then it can be shown that upper and lower limits for the grain size, with less than 1% significance are given by

$$d_{\max}, d_{\min} \approx (\bar{d} / \sqrt{1 + s^2 / \bar{d}^2}) \exp(\pm 3 \sqrt{\log(1 + s^2 / \bar{d}^2)}) \quad \dots (A1.16)$$

This expression is useful in estimating maximum and minimum grain sizes as in the analysis of experimental results in Section A1.3.3.

Distributions of grain size clearly arise during the process because of the thermomechanical processing. A number of programs exist in-house, and have been used in this project, for estimating the evolution of grain size, both recrystallised and unrecrystallised and the fractions recrystallised etc. and for calculating the effect on load. They are based on simple assumptions regarding strain and temperature distributions and are used only as a guide e.g. in the drawing up of schedules. For a detailed analysis of the metallurgical evolution, phenomenological models should be used instead. Those models mentioned above, however, assume that the initial conditions etc are known and neglect, for example, statistical distributions such as mentioned above. They also neglect the random nature of phenomena such as nucleation of grains, (that of porosity in ABAQUS being an exception). In-house software exists for determining the distribution of a quantity functionally related to a quantity such as grain size for which the distribution is given but not when the relationship is more complex e.g. via a set of algebraic or differential equations. The way forward is to develop probabilistic or stochastic models [A1.26] which examine the influence of randomness in process parameters to the overall response of the system.

A1.2.5 Conclusions of the evaluation

1. Use of tabular input in ABAQUS of incomprehensive data for flow stress due to strain rate limitations etc. in the test equipment, can, because of the non-extrapolation in ABAQUS, result in its underestimation during hot rolling simulation and so incur discrepancies in load and torque predictions. Moreover, use of comprehensive data, even if available by sophisticated experimental

techniques, although ideally giving optimal accuracy, would prove cumbersome to use in practice. A suitable constitutive model is therefore preferred.

2. The overstress power law model and the Johnson-Cook analytical flow stress models are available in ABAQUS and are suitable for hardening dominant materials subject to small displacements and strain rate and temperature variations, but not for materials undergoing large deformation and exhibiting softening due to recovery and dynamic recrystallisation.
3. For complex loading conditions in hot rolling etc such as strain reversal and bi-directional loading and changing metallurgical structures e.g. associated with partial recrystallisation, the above-embedded methods in ABAQUS are insufficient. Better solutions for handling complex loading are provided by the incorporation via a UMAT into ABAQUS of viscoplastic and dislocation density based models since these can predict complex history-dependent and time-dependent material behaviour and so are able to trace the strain path and its effect on flow stress.

The various options available at Corus for constitutive modelling with comments evaluating their feasibility are contained in Table A1.2. At the end of this table, the models selected for use in this project and/or further development are itemised.

A1.3 Experimental investigation of complex loading (Task WP2)

A1.3.1 Strain reversal in tension/compression (WP2.2)

Objectives

1. To investigate the effect of alternating tension and compression on stress strain behaviour and recrystallisation kinetics
 2. To compare the results obtained with those in the analogous process of reverse rolling.
- (a) Calibration of the test specimen and the true stress/strain calculation method

To investigate material behaviour of CMnNb steel and 316 stainless steel during strain reversal, it was proposed to use fatigue specimens which would allow compression and tension behaviour being tested in the same test piece. However, the conversion of stroke-force data recorded by the Gleeble simulator into the true stress/strain data proved to be challenging because of the unavoidable barrelling effects in the specimen. Various methods were investigated to convert stroke-force data into true stress/strain data, including 'hot zone' method in which the conversion of data was based on the hot zone in the specimen, 'C-gauge' method in which the variation of specimen diameter was measured directly by a gauge meter, method of extension of gauge length with attempt to reduce the barreling effects, as well as the 'constant volume' method.

The results for extension of specimen gauge length showed that there was no significant difference in true stress/strain data compared with those for the original specimen geometry. The hot zone method which reflects the real deformation region indicated that there was no further improvement. The C-gauge method did not give a reliable measurement in the tension part in compression-tension mode because barreling occurs in compression.

Discussions were held with the University of Oulu on how to minimise the effect. Consequently the fatigue specimens were redesigned, with a more suitable gauge length to diameter ratio. Diagrams and photographs of the specimens employed are shown in Figs. A1.1 and A1.2.

In addition, it was found initially that the resulting true stress/strain curves were significantly different for tension and compression and consequently the results would affect the interpretation of the subsequent stress relaxation data. The problem was solved by imposing an homogenisation treatment on the CMnNb steel plates at 1250°C for 24 hours, which effectively removed segregation existing in the steel plates, from which the fatigue specimens were machined (following scraping of the decarburised material). Consistency of the tension and compression behaviour was then achieved.

(b) Initial strain reversal tests

Tests were carried out to check the suitability of Husain's method [A1.27] for CMnNb steel specimens deformed, at a temperature of 1100°C and a strain rate of 1 s^{-1} , (a) in compression to a strain of 0.2 followed by a tensile strain of 0.2 and (b) to a total strain of 0.4, without reversal. Graphs, for both these cases, of the stress relaxation, the stress-strain variation and recrystallisation behaviour are shown in Figs. A1.3 – A1.5 respectively.

These confirmed that useful information could be extracted using this technique. They are discussed in more detail in Section A1.4 on material characterisation.

(c) Further tests

Further testing was carried out on specimens pre-heated to 1200°C, and held for 15 minutes to dissolve the Nb particles. The schedule employed was that shown in Fig. A1.6 using temperatures of 1000°C, 1050°C and 1100°C. Following deformation, stress relaxation tests were carried out to investigate the recrystallisation kinetics.

Strain reversal tests using CMnNb steel fatigue specimens on the Gleeble simulator were completed. The deformation conditions included 0.4C (compressed to a strain of 0.4), 0.2C+0.2T (compressed to a strain of 0.2 and followed by a tension to a strain of 0.2) and 0.2T+0.2C in each testing temperature 1000°C, 1050°C and 1100°C.

Figures A1.7, A1.8 and A1.9 show the stress-strain curves at the three temperatures for the 0.4 compression tests and strain reversal tests of 0.2T+0.2C and 0.2C+0.2T, respectively. No significant differences in temperature effects were observed but notable Bauschinger effects were evident. Similar conclusions can be drawn from the stress-strain curves shown in for compression followed by tension.

Figures A1.10 and A1.11 compare the results of compression with those of the strain reversal 0.2 C +0.2 T and 0.2 T +0.2 C respectively. Comparison of the first part of the curves demonstrates the repeatability of the test and shows that the compression and tension behaviour is reasonably matched (i.e. corresponding stresses of similar magnitude but opposite in sign).

Figure A1.12 compares the results of the two cases of strain reversal, viz. 0.2T+0.2C and 0.2C+0.2T. It is noticed that only a strain of 0.16 was achieved in the case of 0.2T+0.2C; this may have been caused by the control of the Gleeble machine. It is also observed that 0.2T and |0.2C| match very well and the final stress level is very similar.

A symmetrical cyclic/reversal test was also carried out in order to quantify the Bauschinger effect. The testing temperature was 1100°C and the straining range 0.3. The cyclic data are shown in Fig. A1.13. The Bauschinger effect is not significant in this case; more noticeably, there exists a difference of about 20 MPa in stress level between the steady state cycles and the first two cycles.

The evolution of recrystallisation curves for the above strain paths are shown in Figures A1.14, A1.15 and A1.16 for temperatures of 1000°C, 1050°C and 1100°C, respectively. It is observed that t_{50} (the time corresponding to 50% recrystallised volume fraction) is similar for 0.2T+0.2C and 0.2C+0.2T but somewhat different from 0.4C, suggesting that 0.2T+0.2C as well as 0.2C+0.2T have different recrystallisation kinetics from 0.4C despite having the same equivalent plastic strain.

A1.3.2 Multi-direction compression tests (Task WP2.4)

Objectives

1. To investigate the effects of multidirectional compression on the recrystallisation kinetics
2. To compare with the recrystallisation kinetics of horizontal-vertical (H-V) rolling

Experimental method

Compression testing was carried out on the Gleeble machine at Swinden Technology Centre. Since this model provides compression in a single, fixed, horizontal direction only, deformations could not be applied in different directions simultaneously and so were applied consecutively i.e. by rotating the specimen after the first pass as required. To eliminate effects of specimen geometry, cubic specimens were used because of their symmetry properties. The edge length of the specimens was 10 mm. Compression was carried out on pairs of opposite, horizontal facing faces and the deformed specimen rotated through 90° in a horizontal or principal vertical plane as required, shown schematically in Fig. A1.17. The tests clearly allow several combinations of rotations and non-rotations to be considered, some of these sequences being analogous to those commonly occurring in industrial hot rolling and in particular, horizontal-vertical (H-V) rolling, in which the stock may be turned through 90° in a transverse plane between passes. The aim was thus to establish the differences obtained between turning and not turning the specimen, all other factors including magnitude of the reductions being unchanged.

In order to isolate the effects of strain path pertaining to stock rotation it is important to eliminate extraneous effects particularly those associated with structural heterogeneity and thermal variation. To reduce the effects of the former, the specimens were heated for 30 minutes to coarsen the grain size. In the second case, ideally the specimens should be heated uniformly and constant temperature maintained during testing. To establish complete thermal and mechanical contact after turning barreling should also be minimised. When turning the specimen after the first deformation, this is impossible to achieve on the local model of the Gleeble because of the substantial heat loss from such a small sized test specimen. However, the effects on structure were avoided or, at least, minimised by

quenching the specimen after the first compression. The specimen was then turned, if required, reheated to testing temperature and held at this temperature for 1 minute before carrying out the second compression and re-quenching. To assess the feasibility of the procedure, the microstructures (*respectively quenched after the first compression and reheated for the second compression*) were examined and compared.

The above experimental procedure is shown schematically in Fig. A1.18.

Even if steady and uniform temperature distributions are achieved, the magnitude of any effects of strain path on structure may be dependent on the value of the temperature. An experimental design was adopted in which the effect of temperature was first investigated for a few simple strain paths, followed by a study of a number of sets of complex strain paths characterised by different values of equivalent strain but all associated with the same temperature. This experimental procedure was applied to the compression of Type 316 Stainless steel as described below.

(a) Tests at various temperatures

The tests in the Gleeble compression machine were carried out for low and high temperatures of 900°C and 1100°C respectively and deformed unidirectionally or bidirectionally to strains of 0.3, 0.4 and 0.5 and quenched. Barrelling of the cubes was insignificant.

The microstructures of specimens deformed at 900°C before and after a 90° turn are shown in Figs. A1.19(a) and A1.20(a), respectively and appear to be generally similar, and particularly in the values of the grain size (86 µm and 77 µm, respectively). There is slight difference, however, in the morphology of delta ferrite. Similar conclusions can be drawn from the results for the tests at 1200°C shown in Figs. A1.19(b) and A1.20(b) for both of which the mean grain size was 70 µm.

In examining the microstructures, it was believed that, like the grain size and its distribution, the dislocation densities should also be similar. This claim was not confirmed directly because of difficulties in measuring the latter but was supported by the closeness of the hardness (HV30) measurements viz. 186 and 185, and 182 and 179.

The volume fraction recrystallised was obtained from stress relaxation tests. The test temperatures and the strain were carefully chosen so that the two-volume fraction curves for uni-direction and bi-direction compression at the same temperature could be separated and the variation of the curves with temperature and strain identified. Results of the various uni-direction and bi-direction compression tests are shown in Fig. A1.21 for the selected temperatures and strain i.e. (1100°C, 0.3e), (1050°C, 0.4e), (1025°C, 0.5e) and (1000°C, 0.5e). Differences between the 50% or 95% recrystallisation time are not significant except for the case (1025°C, 0.5e).

(b) Tests for further reductions and sequences of rotations

The next series of multi-directional compression tests reflected, in the main, the various configurational changes of stock occurring in industrial rolling processes such as: turning of bar in transverse plane, edge rolling and cross rolling of plates. More general changes e.g. turning of the whole stock in more than 1 plane were also considered. The latter tests, although not representative of bulk processes in conventional rolling may nevertheless be applicable locally to regions of the stock undergoing complex deformation e.g. near the web flange interface in beam rolling.

The tests were planned to obtain maximum information and to optimise comparability between them and so included;

1. Double hit tests and comparable single hit compressions of the same overall reduction – in a similar manner to the strain reversal tests reported previously.
2. Repeated deformations using different ordered pairs of reductions.
3. Repeated deformations on different pairs of faces, the effect of the order also being investigated (as shown schematically).

A deformation temperature of 1000°C, between the temperatures employed for tests (a) was adopted in the main. The reductions employed of 22.5%, 40% and 53% cover the range of rolling reductions employed in practice and the latter two are mechanically equivalent to a combination of the preceding reductions, in line with objective (1) above.

The tests were carried out as listed according to the programme below. The information given shows the % reduction in height, the occurrence of 90° turns denoted by T and the plane of turning shown by subscripts xz for the horizontal plane and yz for the plane normal to the direction of travel of the anvil. SRX in this context denotes stress relaxation, Q water quench, R reheat to temperature (viz. 1000°C unless indicated otherwise) and M metallographic examination.

- (1) (Heating to 1100°C) 22.5%-->22.5%-->SRX->QM
- (2) 22.5%-->22.5%-->SRX->QM
- (3) 22.5%-->Q->Txy->R->22.5%-->SRX->QM
- (4) 22.5%-->Q->Txy->R->22.5%-->Q->Txz->R->22.5%-->SRX->QM
- (5) 40%-->SRX-->QM
- (6) 40%-->22.5%-->SRX->QM
- (7) 22.5%-->40%-->SRX->QM
- (8) 53.5%-->SRX->QM
- (9) 22.5%-->Q->Txz->R->22.5%-->SRX->QM
- (10) 22.5%-->Q->Txy->R->22.5%-->Q->Txy->R->22.5%-->SRX->QM
- (11) 22.5%-->Q->Txz->R->22.5%-->Q->Txy->R->22.5%-->SRX->QM
- (12) 22.5%-->Q->Txz->R->40%-->SRX->QM
- (13) 40%-->Q->Txz->R->22.5%-->SRX->QM

Comments

Test 1 is similar to a test carried out in stage 1 except that a reduction of 22.5% instead of 20% has been adopted so providing an indication of the sensitivity of the test in terms of reduction. Tests 2 and 3 are also similar to tests in stage 1 except that, in addition, a lower temperature of 1000°C has been employed to show the variation with temperature. Test 5 was set up for comparison with test 2, based on equivalent reductions. Tests 6 and 7 examine the effect of the order of deformation and both are, as above, comparable with test 8. Test 9 along with test 3 investigates the effect of rotation, albeit in different planes, before the second deformation. Test 10 is comparable with Test 4 except that it is a HVH pass. Test 11 is different in that it involves a double hit on the same face and is analogous to cross rolling. Tests 12 and 13, like tests 5 and 6 investigate the effect of the order of deformation, but with a turn in the horizontal plane in between the hits. It can be seen that tests 4, 10 and 11 investigate the effect of 3 hits of equal reduction with turns in different planes applied in the interpasses and the order in which these turns are applied.

Results and discussion

Figure A1.22 shows the decay of load in the stress relaxation test and the corresponding evolution of recrystallisation for the double compression of 22.5% at 1000°C. In this figure the decrease of load and increase in fraction recrystallised with respect to time is shown using both linear and logarithmic time scales, to show respectively, the asymptotic nature of the time variation and the successive stages of primary recovery, dominant recrystallisation and secondary (post) recovery corresponding to the segments of the 'S' type curve. The key parameter generally used is the time to 50% recrystallisation (t_{50}) and, in this case, is equal to 2 s, full recrystallisation taking about 100 s. Figure A1.23 compares the results with those for the test carried out at the smaller strain/pass of 0.25. In this case, recrystallisation is slower in the early stages, 50% recrystallisation taking about 11 s but on comparison of the graphs appears to hasten in the later stages, full recrystallisation taking 100 s. Figure A1.24 compares the results obtained for repeated reductions of 22.5% at temperatures of 1000°C and 1100°C. The recrystallisation appears to be slightly faster at 1100°C, the time for 50% recrystallisation being less than 1.5 s. However, there is a degree of fuzziness in the readings making it impossible to obtain a clearly defined curve and incurring some uncertainty especially in the time for full recrystallisation. Better differentiation has been obtained for the load measurements as shown in Fig. A1.24 (b).

Results for a 90° horizontal turn between applications of the above reductions of 22.5% are given in Fig. A1.25. Comparison with Fig. A1.22 indicates a slower variation in load and recrystallisation, $t_{0.5}$ being equal to 23 s.

When a single compression of 40% is applied, there appears to be little difference in the recrystallisation kinetics from that of the two monotonic reductions of 22.5% as seen on comparison of Figs. A1.22 and A1.26. The kinetics for the single compression is possibly slightly slower with $t_{0.5}$ being slightly greater than 2 s.

When a compression of 40% was followed by a 22.5% reduction in the same direction, the recrystallisation was retarded in the early stages (see Fig. A1.27), the value of $t_{0.5}$ being increased to about 10 s. Reversal of the order of these reductions (cf. Fig. A1.28) makes little difference in the early stages but appears to be faster in the later stages 'full' recrystallisation being attained in about 70 s as compared with 100 s when the greater reduction is applied first. Both these times, however, are less than the time of at least 200 s for a single compression of 40%. Application of the single compression of magnitude 53.5% equivalent to these two compressions, in this case, produces similar recrystallisation

behaviour in the early stages to 22.5% followed by 40%, (cf. Fig. A1.29 viz. a viz. Fig. A1.30) but slightly faster $t_{0.5}$ being 8 s. The process however becomes subsequently slower than both, 'full' recrystallisation being attained at a later time of about 130 s.

The next two figures show the effect of introducing a horizontal turn of 90 degrees between the hits considered in Figs. A1.27 and A1.28. It can be seen that in the first case when the turn follows the larger reduction $t_{0.5}$ is significantly reduced to 1.5 s, but is not reduced as much in the second case when the turn precedes the larger reduction in which case $t_{0.5}$ is equal to 3.5 s. The times for 'full' recrystallisation, viz. 2 min and 1 min however, are only slightly dissimilar from the times for the corresponding sequences without the turns.

Chosen tests involving triple hits were limited by the number of samples available as well as the substantial distortion (e.g. for 3 monotonic hits) that may be incurred. Figure A1.32 shows the results obtained for a triple compression of 22.5% with a 90° horizontal turn at both interpasses (i.e. the case considered in Fig. A1.25 followed by a 90° horizontal turn and further 22.5% reduction in height). The straight graph obtained suggests shallow recovery without recrystallisation. Similar effects are shown in the next two figures corresponding to sequences similar to that above in which one horizontal turn is replaced by a 90° turn in the vertical/transverse plane.

For comparison purposes, the results for fraction recrystallised and load for the full range of strain paths at 1000°C are shown collectively on common sets of axes in Fig. A1.35(a and b) respectively. The results, it can be seen, are divided into four groups in which the behaviour is similar.

Conclusions

1. The results *are consistent with those of similar tests carried out previously and* are qualitatively as expected in terms of changes in strain and testing temperature (minor measurement errors and experimental noise notwithstanding).
2. Increasing the strain (viz. between 22% and 53.5%) reduces the time for 'full' recrystallisation.
3. A greater time for full recrystallisation appears frequently to be associated with a smaller time to 50% recrystallisation (and vice versa).
4. Following a hit by a further hit (cf. 40% and 40% followed by 22.5%) appears to retard recrystallisation but overall is faster.
5. Reversing the order of 2 monotonic reductions makes very small absolute differences in the time for 50% recrystallisation and small relative changes in the time for full recrystallisation.
6. Replacing a large single hit by two hits of equivalent magnitude appears to produce slightly slower early recrystallisation kinetics, (cf. 53.5% into 40% and 22.5%) but overall is faster.
7. A 90° turn between 2 passes in conjunction with an increased reduction in either pass (cf. 40% from to 22.5%) tends to hasten early REX but without the large reduction significantly retards recrystallisation.

8. Triple hits of the magnitudes considered above appear to prevent recrystallisation irrespective of the sequence of the pairs of planes on which they are applied i.e. this arises basically because the reductions of 22.5% are not sufficiently large (cf. points 2,4 and 7).

A1.3.3 Stock rotation during rolling

A1.3.3.1 Horizontal – vertical rolling (Task WP2.6a)

Objectives

To assess the effects on microstructure (i.e. fraction recrystallised and recrystallised grain size) and load, during the experimental hot rolling of Type 316 Stainless steel bars to reductions typical those employed on plant, of (a) interpass time (b) interpass turn through 90° and (c) the heating (inc. reheating) strategy.

To compare the results with analogous results obtained from bi-compression techniques on the Gleeble.

Associated requirements

In order to meet the above objectives the following criteria need to be met.

- (i) Similar specimen thicknesses, and reductions and temperatures common to those used in the compression testing should be employed.
- (ii) The thermomechanical conditions should be typical of those employed on the plant. In particular, a key factor for simulation of rolling processes, not included in the compression tests, is the interpass time. A range of times compatible with the other conditions should therefore be used.
- (iii) Specimen geometry needs to be carefully controlled throughout the process.
- (iv) Operating parameters such as load and torque need to be accurately measured and maintained within the mill limits.
- (v) The metallurgical state, in particular, the grain size and the fraction recrystallised needs to be clearly characterised at each stage of the process.
- (vi) The temperatures need to be monitored and controlled during heating and cooling.

Methodology for construction of rolling trials

Rolling schedules were drawn up for initial feedstock as supplied i.e. nominally 23 mm square and approximately 20 cm long, the length sometimes being cut to create more specimens. Initial designs were based on experience using moderately large reductions to highlight any significant strain path effects, (of the order of 20% used in the compression test). The schedules were then corrected iteratively as necessary so that they satisfy the criteria above using the various models listed in section A1.2 and the experimental facilities mentioned in section A1.4 Initial stock temperatures in the range 1100°C to 1150°C and interpass times up to 30 s were considered. For detectability, grain sizes in the range 50 – 120 µm were specified.

Solution of the Ekelund spread Equation (A1.2) implemented using MATHCAD together with application of inhouse software (cf. Section A1.2) suggested that the 23 mm square bar reduced by 20% would have an exit width of at most 24.5 mm and, if then reduced by further 20% without turning, to a width of 25.8 mm. If turned between passes, the corresponding final thickness would be 19.6 mm and the estimated width 21 mm.

Temperature distributions in the specimen were obtained using an ABAQUS FE heat transfer model for initial soak temperatures of 1150°C and 1120°C. Cooling curves at various locations, obtained from the predictions, are shown in Figs. A1.36 and A1.37. From the graphs it could be seen that there would be a rapid decay in temperature at the stock faces, especially for the higher temperature. Orowan's formula [A1.1] and the inhouse software using an initial grain size of 50 μm predicted loads in the range 75 to 81 kN for pass 1 and 81 to 89 kN for pass 2 without turning; when the stock is turned the load predicted by Orowan's formula was of the order of 73 kN (no predictions from additional inhouse software were possible in this case since turning option is not included, the loads for partial recrystallisation would be slightly higher). From the predictions, even taking into account adjustments for partial recrystallisation, the load was expected not to exceed the operating limit of 10T for the Cavendish mill. The predicted torque of at most 1.4kNm was also considered not to present any problems.

For grain sizes at the upper end of the range, predictions of fraction recrystallised, using the Barbosa formula, [A1.12] were very small, viz. of the order of 3%. For grain sizes at the lower end, corresponding predictions are of the order of 18% but the recrystallised grain sizes differs little from the initial value and are small. Predictions of the microstructural evolution were also obtained using the inhouse software, with a grain size of 50 μm and employing a low emissivity corresponding to cooling the specimen isothermally i.e. without the effect of radiation, for the 3 interpass times of 10 s, 20 s and 30 s.

After a 10 s interpass time, the predicted temperature decreases from 1097°C at the centre to 1083°C at the surface; the predicted fraction recrystallised ranges from 29.5 to 33.3% and the recrystallised grain size from 42.1 to 42.7 μm . The predicted load in pass 2 for these temperatures and accumulated strains of up to 0.494 together with the above degree of partial recrystallisation, is 74.6 kN. The corresponding torque is 1.2 kN m.

For the 20 s interpass time, the predicted temperature distribution is little different from that for 10 s but the fraction recrystallised and grain size increase and lie in the range 47.2 to 52.4% and 46.7 to 47.9 μm respectively. The predicted load and torque for pass 2 are 75.8 kN and 1.2 kN m respectively.

For 30 s interpass time, the fraction recrystallised and recrystallised grain size lie in the ranges 58.1 to 63.7% and 52.0 to 53.8 μm respectively. The predicted load is 77.1 kN and the torque is 1.2 kN m.

All the loads predicted, even for partial recrystallisation, are less than the 10 t operating limit.

Initial specification of trials

From the above analysis, the following schedules were therefore selected for initial experimental investigation:

- Dimensions: 23 mm square cross section, approximately 20 cm long
- Grain Size: 50 μm

- Reheating temperature: 1115°C
- Initial Stock Temperature: 1100°C
- No. of Passes: 2
- Time for start of first pass: 6 s
- Nominal Reductions: 20% per pass
- Rolling speed: 50 rev.min⁻¹
- Mean strain rate: 5.4-5.9 s⁻¹
- Interpass times: 10 s, 20 s
- Rolling schedules:
 - Schedule 1 (horizontal-horizontal rolling):
 - Pass 1: Draughting from 23 mm to 18.5 mm (actual reduction 19.6%)
 - Pass 2: Draughting in the same direction to 15 mm (actual reduction 18.9%).
 - Schedule 2 (horizontal-vertical rolling):
 - Pass 1: Draughting from 23 mm to 18.5 mm (actual reduction 19.6%).
 - Turn 90° in transverse-vertical plane.
 - Pass 2: Draughting from nominally 24.5 mm (to be checked by measurement between passes 1 and 2) and reduced by 20%.

Experimental Method

As noted in Section A1.4, the average grain size in the unprocessed samples is of the order of 10 µm. To maximise detectability of changes in the microstructure, the samples needed to be heat treated to induce grain growth. To find the holding time required, a graph (shown in Fig. A1.38) of grain size versus time at this temperature of 1100°C, obtained from experimental data by means of the TableCurve2D package was used. A suitable range of grain sizes (viz. 40 – 60 µm) was therefore selected, significantly larger than 10 µm but on a portion of the curve in which the growth rate is low and hence controllable (i.e. the grain size should not change appreciably whilst the stock is about to exit the furnace prior to rolling). From the graph it can be seen that this requires a hold time of 20 min at 1100°C.

For ‘continuous’ measurement of rolling load and torques for both rolls, the mill was instrumented with load cells and the spindles with strain gauges and slip rings. The signals were transmitted at regular small time intervals to a laptop pc, where the data were recorded and from which traces of the variables could be obtained. The traces were used for determining the onset and duration of steady state conditions, and for confirming subsequently the interpass times taken whilst the trial was in progress. In order to capture the metallurgical structure for subsequent examination, the specimen was placed promptly, on exit from the final pass, into a bucket containing water at ambient temperature, which was agitated to optimise the heat lost by the specimen.

The first specimen was removed from the furnace, rolled in a single pass and quenched immediately. Traces of load and torque are shown in Fig A1.39 from which, the start and duration of rolling were estimated viz. at about 0.4s for 0.6 s and steady state conditions were ascertained. The measured thickness was in the range 18.7 to 18.8 mm i.e. slightly higher than set. The steady state exit width was 25.7 mm. Equating this to the subsequent entry thickness in the second pass following an interpass turn of 90°, the final thickness expected after 20% reduction is 20.56 mm.

On exit from the rolls, the specimen was quenched. Two transverse sections were later taken from the rolled bar and the metallurgical structure observed by quantitative metallography. Photographs of samples taken at the $\frac{1}{4}$ depth position and in the central region are shown in Fig. A1.40(a and b) and Fig. A1.40(c and d) respectively. Values of the average grain size were equal, to the nearest micron, to 38 μm . However, from inspection of the figures, it can be seen that there is a wide variation in grain size throughout the samples (standard deviations 3 μm and 6 μm), the value tending to be smaller in the central regions. Moreover, the wide variation including low grain size and also the non-equiaxed nature of the grains found in such figures as above made planned determination of the strain distribution infeasible.

A second specimen was removed from the furnace and rolled in two monotonic passes with an interpass time of the order of 10s and quenched on exit from the final pass. Traces of the load and torque for each pass are shown in Fig. A1.41(a and b). To deduce the potential evolution of recrystallisation on exit from the rolls, initially only three 12 mm thick sections, transverse to the rolling direction, were taken upstream from a position 70 mm from the leading end. One section (A) was allocated for observation of the structure on exit from the rolls, the remaining two (B, C) for determination of the structure (i) shortly after exit and (ii) when recrystallisation was almost complete i.e. at the essential end points of the recrystallisation curve. The information from (i) when compared with that obtained from section A was useful in determining the initial gradient of the recrystallisation curve. Times of 5 s and of 60 s were chosen respectively for (i) from practical considerations and for (ii) estimated from theoretical recrystallisation curves. With a view to attaining these states, the specimens were placed beneath the platens of the Gleeble, heated rapidly to 1100°C and then held at temperature respectively for the times specified. Samples were taken at the midwidth positions for metallographic examination and measurement of the fraction recrystallised.

Photographs of the observed microstructure are shown in Fig A1.42 (a-f), and values found for the average grain size, together with the associated standard deviation in Table A1.3.

A third specimen was similarly heated and rolled in two passes with an intermediate 90° turn, and an interpass time of the order of 10 s as above and quenched on exit from the final pass. Traces of the load and torque for each pass are shown in Fig. A1.43(a and b). A similar procedure to that for the second specimen was followed to reveal the potential recrystallisation kinetics, viz. by taking initially three 12 mm thick sections, from a position 50 mm from the leading end. Unlike the previous case, these sections were approximately 21 mm square. Following heating to 1100°C and holding for 5 s and 60 s, two samples were taken at the mid-height and mid-width of the section to investigate the symmetry or otherwise of the metallurgical structure. Photographs of the observed microstructure are shown in Fig. A1.44(a-e) and values found for the average grain size, together with the associated standard deviation are given in Table A1.4.

Discussion of results and issues arising

From the results the following observations were made:

1st Specimen:

1. The loads obtained are comfortably within the 10 t operating limit and are reasonably balanced between the open and drive side.
2. There is an asymmetry between the separate roll torques, the value for the bottom roll being greater.

3. The measured exit thickness is slightly greater (by about 0.2-0.3 mm) than the specified exit thickness. This is probably due to mill spring which was not taken into account in these trials.
4. The measured grain sizes (on average 38 μm but with a wide variation throughout the samples, with respective standard deviations of 3 μm and 6 μm), are at the lower end or even below the range specified even though it is unlikely that recrystallisation has occurred by this stage.

The low growth of the grains is likely to be related to the temperature distribution within the specimen or its variation throughout the trial. In particular, a sufficiently high grain size may not have been attained either because the region of the stock did not attain the required temperature or, otherwise, was not held at temperature for a sufficiently long period. Temperatures were not measured for this particular trial but predictions at specific locations within the specimen and at selected times prior to rolling were obtained using an ABAQUS FE heat transfer model assuming an initial uniform temperature of 1120°C. From the predictions, cooling curves have been obtained. The graphs indicate a rapid decay in temperature initially at the surface and subsequently in the central region. The short period prior to, and during rolling, of this trial, however, is unlikely to have any significant effect, suggesting that no portion of the stock had maintained, for sufficiently long, the required temperature of 1100°C (or even attained it).

On the basis of these predictions, a further trial was carried out to monitor the specimen temperature during reheating and subsequent cooling on exit. The evolution of the measured temperature at the centre and surface of the stock for these stages is shown in Fig. A1.45(a and b). The absence of a plateau in the first figure confirms the lack of a holding period and the low first recorded temperatures in the second graph tend to confirm that the required temperature may not even have been attained.

2nd Specimen:

The results observed indicate that:

1. For the first pass the loads are balanced and within operating limits and the torques somewhat asymmetric, replicating the observations for the single pass trial above.
2. The required interpass time of 10 s was attained.
3. In the second pass the loads are higher than those in the first pass and approach the operating limit.
4. The longer times for this trial and consequent temperature losses indicated from the predictions in Fig. A1.37 suggest that the trial did not take place under isothermal rolling conditions.
5. This temperature decay, coupled with the low initial temperature and grain size, according to the equations of the recrystallisation kinetics A1.7 –A1.11) [A1.11], would give rise to low fractions recrystallised and recrystallised grain sizes, difficult to detect.

The high grain size of 61.9 μm in the unreheated sample at the quarter depth position (where the measurements after the single pass were taken) could be taken as a confirmation that grains had grown to the required size. This is consistent with the observed grain size of 66.5 μm observed after 5 s. The moderate value of fraction recrystallised (59%) could be due to experimental error. The low grain size of 36.7 μm found after 60 s is difficult to explain.

3rd specimen:

1. The loads and torques are slightly higher than those in the nominally identical pass for the previous case, also unsteady and slightly less balanced, indicative of 'noise'.
2. The major difference, if path dependence is important, should occur in the second pass following the rotation. Noticeable differences do occur, the loads in this case being smaller and steadier on the drive side but greater and more variable on the open side. The torque for the current case tends to be smaller on the top but greater on the bottom shaft. The asymmetry, occurring mainly in the torques could, in principle, be caused either by different conditions between top and bottom rolls or by inhomogeneities in the specimen. The possibility of significant geometrical inhomogeneity was considered and ruled out; the evidence suggests that the effect is likely arise from initial inhomogeneities in the material or changes in the metallurgical structure associated with the rolling process.

From inspection of Table A1.4 similar qualitative remarks to those above may be made concerning the structural changes incurred by the rolling with the 90° turn viz. recrystallisation is initiated probably resulting in a larger recrystallised grain size which appears to grow with time. The results are however respectively smaller and greater than the corresponding values in Table A1.3. More precisely, the material in location 1, which has been in contact with rolls once following the turn is recrystallised less than the material, which has been in contact with the rolls twice. This is consistent with the observations for bi-directional compression and tensile-compressive tests. For the material in location 2, especially at the centre, the recrystallisation is greater because the material undergoes repeated compressive strains, which from the geometry of the stock are probably greater than those at the surface for the H-H rolling. The lower loads noted above for H-V rolling suggests that the dominant effects are probably those such as in location 2 i.e. that the recrystallisation is greater when turning is introduced. Further tests were recommended to confirm these effects and to investigate also the effects of interpass time and temperature variation.

Issues

Before carrying out further trials, a number of issues arose relating to (a) uniform attainment of the required initial grain size and reheating temperature (b) minimisation of radiative heat loss in specimen in transit between furnace and rolls and between passes (c) setting the correct roll gap (d) extraction/presentation of meaningful data on grain size, strain distribution and REX.

Addressing the issues

1. Uniform attainment of the required grain size throughout the specimen was problematic because of the presence of secondary phases in the as-supplied sample as noted in Section A1.4. It was decided therefore that two reheatings be carried out viz. (a) a heat treatment prior to the trial, under predetermined and controlled conditions aimed at minimising the heterogeneities in the structure and (b) one to attain the required initial grain size prior to rolling. Measurements would be taken in the region of the specimen containing the primary phase.
2. Attainment of a uniform temperature distribution in a sample has been found to be difficult, if not impossible, to achieve in practice. As predicted in [A1.11] there is a gradient near the boundary during the heating of Gleeble specimens, as also suggested in Figs. A1.36 and A1.37. The heating conditions for attainment of approximately uniform temperature were therefore found experimentally by monitoring the temperature readings from an inserted thermocouple.
3. From theoretical curves of recrystallisation, for larger interpass times around 30 s (cf. Fig. A1.37), it can be seen that in this case there is a large temperature drop so that, unlike the 10 s case, approximately isothermal conditions cannot be assumed. Various methods of insulating the specimen were therefore examined including immersing the specimen or placing in a suitably heated, specially designed, refractory unit. These measures were ruled out as they would probably not reduce the heat loss significantly or would be impractical. It was therefore decided to use, instead, an isothermally based equivalent time theoretically determined from the recrystallisation equation as outlined in Section A1.2 in Equations (A1.13-A1.15).
4. To determine the roll gap required, taking into account mill spring, a spring curve for the mill was obtained by rolling a number of specimens to various gauges, reading the steady state load from the trace and determining the deviation from the set gauge. It is shown in Fig. A1.46. The gauge setting for each of the passes has been obtained by subtracting the spring for the loads noted previously without any corrective iteration.
5. To obtain meaningful information on microstructure it is important not only to determine the microstructural evolution after the final pass but also the structure prior to the first pass and that between passes. To fulfil the latter requirement, a specimen was quenched after the first pass and the structure determined.

In order to make optimal use of the specimens, the evolution of fraction recrystallised and recrystallised grain size was not measured by rolling of several specimens, quenched at different times, but by quenching, for each case, a single specimen approximately 2 s after rolling and measuring the REX characteristics in reheated sections of this specimen after various times of cooling. This method assumes that the strain energy trapped in the quenched section will resume the recrystallisation, as uninterrupted, if reheated to the appropriate temperature. The information is presented as curves of REX percentage v log time.

Series 2 of rolling trials: Effect of interpass time and temperature

Programme

Before rolling, the remaining specimens of Type 316 stainless steel were homogenised by heating them to 1100°C in the heat treatment furnace, holding at temperature for 2 hours and allowing to cool in air. Samples from the ends of the specimen were then examined metallographically and measurements taken of the grain size. Micrographs are shown in Fig. A1.47 and values of the mean grain size and standard deviation are given in Table A1.5. Compared with the small values measured in specimens used in tests previously, the sizes are unexpectedly high at around 90 microns with standard deviations of the order of 10 microns. A moderate degree of heterogeneity therefore still remains as can be seen in Fig. A1.47. It was expected that the heterogeneity would decrease towards the mid-length position of the bar but samples were not taken here as it was thought at this stage of the procedure that there would be a significant deviation from steady state conditions during rolling in specimens of half the length (c. 10 cm) used previously.

To avoid the problems regarding control of temperature during the interpass, it was decided therefore to adopt the approach used in the triaxial compression tests of quenching the specimen immediately after each pass and reheating to the temperature required.

Two alternative scenarios, I and II, were considered in the trials. In scenario I, to generate moderate recrystallisation, a trial was planned in which a bar was rolled in pass one with the same reduction as above, reheated to 1040°C but then held for 60 s only, quenched and reheated to 950°C, before rolling in pass two by a 20% reduction. The two respective heating profiles are shown in Figs. A1.48 and A1.49. As predicted from the model equations, further recrystallisation at this temperature, if quickly attained, should be minimal. In scenario II, to investigate the effect of zero or negligible recrystallisation on subsequent loads and torques and compare the evolved grain structures, a similar schedule but with the omission, after pass one, of the reheating to 1040°C was also drawn up. In both these scenarios, the effect of strain path is examined, according to the remit in the technical annex, in relation to the introduction or otherwise of interpass turns. The bars employed in the two scenarios were labelled respectively 4R1F64A and 4R1F284A, with subcodes 1 and 2 corresponding to no turn or turn as defined in Table A1.6.

The temperature in the trials was also carefully monitored using a thermocouple inserted into the specimen; if the temperature required was not attained the PID controller of furnace temperature was reset. The mill, including the instrumentation, was also checked and the spurious torque oscillations noticed previously eliminated. The spring curve for the mill was employed to determine the deviation from the set gauge. For the first pass a spring of 0.49 mm was noted.

The traces of load and torque are given in Figs. A1.50 and A1.51 for bars 4R1F284A (1 and 2) and in Fig. A1.52 for bar 4R1F64A (1). After rolling the bar was quenched immediately in order to capture the actual recrystallised structure. Samples taken from the end of the bar after the first quench in the schedule were used to investigate the recrystallised structure and the effect on loads and torques on subsequent rolling in which the bar was rolled with a gap setting, allowing for spring back, of 14.0 mm. Further samples were taken in tests to determine the grain sizes at the end of rolling and the subsequent structural evolution. The samples consisted of a 10 mm thick slices, labelled F64A1-1 etc., one cut from each bar at the end furthest from the thermocouple and of a number of 10 mm cube samples, F64A1a, F64A1b. The 10 mm thick slices underwent metallographic examination. The remaining samples were first heated to 950°C beneath the platens of the

Gleeble and held at temperature generally for 10 s and 60 s respectively, quenched and then examined. Micrographs of the structure at these times for the 4 cases are shown in Figs. A1.53 - A1.56 respectively. In each figure, the structure at the $\frac{1}{4}$ depth and centre positions is shown for the above times. Average grain sizes with standard deviations were obtained for bars 4R1F284A2, 4R1F64A1 and 4R1F64A2 and are given in Tables A1.7 – A1.9. Grains in samples from bar 4R1F284A1, which did not have the extra reheat, exhibited significant variability, making it impossible to give measurements of their size. Values were therefore in randomly chosen individual grains along the centreline of the sample at positions 1 mm, 6 mm from the edge where the microstructures were predominantly equiaxed, and at 12 mm where the grain size was mixed and sometimes elongated. Measurements in μm , in this case, are expressed as 'feret max.' (the largest dimension regardless of orientation) and 'feret min.' (the maximum dimension perpendicular to 'feret max.') and are shown in (Table A1.10).

Direct measurement of recrystallisation by optical metallography was not attempted as it was considered unreliable for these specimens.

Discussion of results

(a) Effect of temperature

On comparison of Figs. A1. 52(a) with A1.50(a), for rolling with and without the reheat to 1040°C, the entire profiles are very similar suggesting, surprisingly, that recrystallisation, if taken place, does not affect the magnitude of the load. Differences do exist however between the corresponding torque traces. A qualitative assessment of recrystallisation was made using the grain size results obtained by optical metallography. From Fig. A1.55, there is little evidence of widespread recrystallisation in the bar having had the extra reheat to 1040°C, the small changes in average grain size over the time period being observed (viz. to 55, 55.7 and 55.8 microns probably being due to normal grain growth. However, close inspection of the figure reveals a small network of grains at the centre, suggesting recrystallisation was likely to have occurred locally but not quantifiable with reasonable accuracy.

For the bar not having had the extra reheat, the values cannot be used to accurately determine the average grain size, but do provide some indication of the size of the largest grains present viz. 217 to 531 μm in the unreheated sample after rolling and 250 to 441 μm , 204 to 698 μm for the samples heated to 950°C and held for 10 s and 60 s. They can therefore be compared with estimates of the maximum grain sizes for the previous case. Assuming that the grain sizes are lognormally distributed, the maximum grain size can be estimated using equation A1.16, giving results of 71.1, 63.3 and 64.0 μm . (*If the grain sizes were normally distributed application of the '3 standard deviations rule' would give grain sizes of 69.4, 62.9 and 63.6 microns.*) It is clear that the large difference between the values even for the smallest measurement of 204 μm cannot be explained simply as the difference between recrystallised grains and grown grains; it is more likely to reflect inhomogeneities and present in the material in its initial state and variability between specimens. The large grains at this position also showed some directionality with the elongation tending to lie parallel to the top/bottom surface.

The presence of smaller networks of grains throughout the specimens suggested extensive recrystallisation. The anomaly between the above two sets of results, in which indications of the absence or presence of recrystallisation are contrary to expectation, is difficult to explain. It was decided therefore to discern whether recrystallised structures are present by the use of EBSD techniques for both the above cases, as described in more detail in Section A1.4.

This work was carried out using TSL EBSD equipment fitted to a Philips XL30 SEM in accordance with Work Instruction CPR SUS 324. Figures A1.57 and A1.58 provide the IQ (Image Quality) Maps for specimens 4R1F64A1B and 4R1F284A1B respectively for the cases with and without the extra reheat. However, the quality of the EBSD patterns obtained was not fine enough to enable indexation, by the software, of sufficient points (the white areas on IPF maps) to detect regions of recrystallisation.

(b) Effect of turning

1. Through 90° in transverse-vertical plane.

To investigate the effect of changes in stock configuration (at constant reduction) on load and recrystallisation, a similar test to that above was carried out except that the stock was turned through 90° in the transverse vertical plane before the second pass. The loads and torques, are shown in Fig. A1.52(a and b), for the bar in the pass after a 90° turn are smaller than those in the previous figure for the specimen without the turn. This result is qualitatively in agreement with predictions obtained using the Orowan load model [A1.1] and is associated with the changes in stock geometry. It also suggests less strain energy than in the previous case for subsequent promotion of recrystallisation. To compare the microstructure, tests analogous to those for the rolling without the interpass turn were carried out. For the specimen rolled with the interpass turn, grain sizes obtained from the micrographs corresponding to the above values of 55, 55.7 and 55.8 microns are respectively 78.2, 76.0 and 82.5 microns suggesting that the joint effect of reducing the reheating temperature and turning the stock increases the grain size which is consistent with absent or possibly retarded recrystallisation or grain growth. It would be useful to factor out the effect of the reheating temperature and consider only the effect of path change. In principle, this could be done by first estimating for this case the maximum grain sizes as above giving values of 108.8, 99.7 and 103.0 microns (104.6, 97, 101.1 microns using the normal distribution), and then comparing with the above values of 217, 250 and 204 microns. The differences are more supportive of an effect of rotation on retardation of grain refinement but, are still too large to attach any great significance to the results.

2. Turning in a horizontal plane

Rolling trials have also been carried out for instance to investigate the effect of turning in the horizontal plane (cross-rolling) on the evolution of microstructure. In particular, the stock followed a similar processing history to 4R1F64A1 above viz. was rolled at 1040°C from 23.1 mm to 18 mm, was then quenched, reheated to 1040°C, held at temperature for 60 s, quenched and reheated to 950°C, but then turned through ~ 45° in the horizontal plane before rolling in the second pass to 14 mm. On comparison with the results for specimen 4R1F64A1, the average grain size is smaller for the two subsequent times. However the differences are not sufficiently significant to support the hypothesis that turning retards the grain refinement. Indeed the mixed grain structure shown in Fig. A1.56 of this set does not provide any clear indication of recrystallisation. The results also suggest that the grain size does not evolve significantly as the holding time is increased.

Conclusions

1. Anomalies have been detected concerning the effect of temperature on recrystallisation, indicating the opposite of the outcome generally expected. This may be indicative of the structural variation within and between samples and insufficient control of heating and cooling schedules.

2. There is slight evidence that changing the strain path through interpass turns retards evolution of recrystallisation and reduces grain refinement. The evidence, from the limited number of trials possible, therefore is not sufficiently significant to establish whether this is a secondary effect or even a real effect for the steel quality studied.

A1.3.3.2 Forward/Reverse plate rolling (Task WP2.6b)

The objectives of this task are: -

1. To investigate the effects of reversal rolling on the recrystallisation kinetics.
2. To use the results for the benchmark exercise

The reversal rolling trials were carried out on the 3 high laboratory plate mill at Corus Swinden Technology Centre. Figure A1.59 shows the mill facilities used in this rolling trial.

Two schedules were used for rolling temperatures of 1020°C and 1050°C consisting of the following steps viz.

Schedule 1

- Reheat 1250°C for 30 min to achieve a grain size of about 150 µm and to dissolve the Nb particles.
- Soak in the furnace at 1120°C to achieve as uniform as possible grain structure and temperature.
- Roll at 1020°C at fixed speed of 78 rev/min
- Roll without reversal to following sequence of stock heights:
 - Pass 1 (Normal direction): 30 mm to 25 mm, interpass time=10 s (manually change rolling direction)
 - Pass 2: 25 mm to 22 mm, interpass time = 12 s
 - Pass 3: 22 mm to 19 mm, interpass time = 12 s
 - Water quench
- Roll with reversal to following sequence:
 - Pass 1(Normal direction): 30 mm to 25 mm, interpass time = 10 s
 - Pass 2 (Reverse direction): 25 mm to 22 mm, interpass time = 12 s
 - Pass 3 (Reverse direction): 22 mm to 19 mm, interpass time = 12 s
 - Water quench

Schedule 2

- Reheat at 1250°C for 30 min to achieve a grain size of about 150 µm and to dissolve the Nb particles.
- Soak in the furnace at 1140°C to achieve as uniform as possible grain structure and temperature.

- Roll at 1050°C at fixed speed of 78 rev/min.
- Roll without reversal to following sequence of stock heights:
 - Pass 1 (Normal direction): 30 mm to 25 mm, interpass time=10 s (manually change rolling direction)
 - Pass 2: 25 mm to 22 mm, interpass time = 12 s
 - Pass 3: 22 mm to 19 mm, interpass time = 20 s
 - Water quench
- Roll with reversal to following sequence:
 - Pass 1 (Normal direction): 30 mm to 25 mm, interpass time = 10 s
 - Pass 2 (Reverse direction): 25 mm to 22 mm, interpass time = 12 s
 - Pass 3 (Reverse direction): 22 mm to 19 mm, interpass time = 20 s
 - Water quench

Four dummy trials were performed to check the gauge and the roll gap and to get the control and recording system working correctly. The rolled profiles are shown in Fig. A1.60.

Eight formal trials have been carried out with details as follows:-

- TRIAL 1: Tre=1250°C, Tsoak=1120°C, Troll=1020°C, 3 passes in which the original thickness of 30 mm was reduced to 25 mm, 22 mm and 19 mm, respectively, without reversal rolling.
- TRIAL 2: Tre=1250°C, Tsoak=1120°C, Troll=1020°C, 3 passes in which the original thickness of 30 mm was reduced to 25 mm, 22 mm and 19 mm, respectively, with reversal rolling.
- TRIAL 3: Tre=1250°C, Tsoak=1120°C, Troll=1020°C, 3 passes in which the original thickness of 30 mm was reduced to 25 mm, 22 mm and 19 mm, respectively, with reversal rolling. REPEAT of TRIAL2.
- TRIAL 4: Tre=1250°C, Tsoak=1120°C, Troll=1020°C, 3 passes in which the original thickness of 30 mm was reduced to 25 mm, 22 mm and 19 mm, respectively, with reversal rolling. REPEAT of TRIAL2.
- TRIAL 5: Tre=1250°C, Tsoak=1140°C, Troll T=1050°C, 3 passes in which the original thickness of 30 mm was reduced to 25 mm, 22 mm and 19 mm, respectively, without reversal rolling.
- TRIAL 6: Tre=1250°C, Tsoak=1140°C, Troll T=1050°C, 3 passes in which the original thickness of 30 mm was reduced to 25 mm, 22 mm and 19 mm, respectively, with reversal rolling.
- TRIAL 7: Tre=1250°C, Tsoak=1140°C, Troll T=1050°C, 2 passes in which the original thickness of 30 mm was reduced to 25 mm and 22 mm, respectively, without reversal rolling.

- TRIAL 8: $T_{re}=1250^{\circ}\text{C}$, $T_{soak}=1140^{\circ}\text{C}$, $T_{roll}=1050^{\circ}\text{C}$, 2 passes in which the original thickness of 30 mm was reduced to 25 mm and 22 mm, respectively, with reversal rolling.

(T_{re} in the above represents the reheating temperature, T_{soak} the soaking temperature, T_{roll} the rolling temperature),

Experimental results

Operating parameters

Figure A1.61 shows examples of torque and load in Trial 4 and Fig. A1.62 for Trial 1. The averaged values can be obtained in order to compare with those predicted from metallurgical models.

It can be seen that the loads for Trial 1 (without reversal) are approximately 35, 34, and 38 tonnes for the first, second and third pass, and the loads for Trial 4 (with reversal) are approximately 32.5, 34.5, and 38.5 tonnes. However, it is not certain if these differences were caused by the reverse rolling or by other variables such as interpass time, rolling temperature that are difficult to control in the rolling trials.

A1.3.4 High speed rolling

(See also report of TU Freiberg contribution in Appendix 5)

Objectives

1. To investigate the effects of Horizontal-Vertical rolling on REX kinetics.
2. To investigate the effects of high speed rolling on flow stress data.
3. To examine the effectiveness of extrapolation of the constitutive models with higher strain rates.

This work has involved collaboration between Corus and TU Freiberg, using the high speed rolling mill of the latter. Trials on the four stand mills were scheduled by Corus originally for 900°C , 1050°C and 1200°C at 40 m/s with a conductive heating temperature of 1250°C to achieve an initial grain size of 180 μm . The low speed (5 m/s) rolling was also planned to calibrate the constitutive models. Details of the trials are contained in the contribution of TU Freiberg in Appendix 5.

FE modeling, taking into account cooling, of the Freiberg two stand rolling was carried out by Corus using ABAQUS. Details are given in Section A1.7.

A cooling model was created with a furnace discharge temperature 1250°C . The variation of temperature with time was examined to provide an understanding of the temperature evolution during rolling. The FE model showed that the strain distribution over the cross section of rolled rod was non-uniform, being highest at the centre (0.7) then the middle (0.55) and finally the surface (0.3). So if complete REX occurs in the centre, there may be partial REX at the surface.

A1.4 Material characterisation (Task WP3)

A1.4.1 Facilities and techniques

Of the techniques for material characterisation mentioned in the main section of the report and described in Ref. [A1.28], those employed by Corus in this project are optical metallography, electron back scattering diffraction (EBSD), stress relaxation and, occasionally, the double hit technique. Supporting facilities for reheating, sectioning, machining, etching and polishing etc are readily available and have been utilised. For stress relaxation, the local Gleeble has been used to deform and hold the specimen at a given strain. Specimens have also been held and reheated between the anvils of the Gleeble for rapid attainment of high temperatures with minimal recrystallisation,

A1.4.2 Background: Material allocation

Of the total set of steel qualities under overall investigation, contained in the template given in Appendix 7, those for which Corus is responsible are CMn, CMnNb steels and Type 316 Stainless Steel.

CMnNb steel was originally planned to be the main material for rolling and laboratory testing since it is widely used in Corus, and is readily available. CMnNb steel was therefore allocated for use, (a) by Corus, in tensile/compression tests on the Gleeble and in laboratory rolling, (b) by CEIT in torsion testing, and (c) by Freiberg in forward-reverse rolling. Eighty CMnNb steel plates (30 mm in thickness, 120 mm in width, 400 mm in length) were acquired by Corus, thirty were sent to Freiberg for use in their high-speed rod mill and five to CEIT for use in their torsion machine. The composition is given in Table A1.11, has a carbon equivalent value of 0.39 and is similar to that for the steel used by the University of Oulu in its tension/compression testing, thus readily generating data for comparison purposes.

Type 316L stainless steel was procured for use in the Gleeble multi-direction tests and the laboratory horizontal-vertical rolling trials. Its composition is given in Table A1.12. Although not as widely used as in the past when stainless strip was rolled by Corus, it is regarded, in common with microalloyed steels, as a material for which strain path effects are important and ample stocks were available.

The CMn and stainless steels are analogous to, i.e. can be compared and contrasted with, the medium C and duplex stainless steels investigated by CSM, the University of Oulu and CEIT.

A1.4.3 Initial observations on material characterisation

- (a) A brief assessment on the material characterisation of CMn steels is given in Reference [A1.28]. In particular, it has been found that optical metallography is difficult if the material undergoes a phase transformation during quenching and ambiguous results may be obtained if etching is carried out on prior austenite grains. For stress relaxation and double hit techniques, although recrystallisation occurs simultaneously, it occurs at a faster rate in the former. While the EBSD technique is generally regarded as being the most accurate and powerful technique, it is only possible if no phase transformation occurs after recrystallisation.
- (b) CMnNb steel recrystallises moderately quickly, the speed being retarded by the presence of Nb. However, since it is transformable, definition of the deformed

austenite grains can prove extremely difficult following transformation to ferrite, pearlite or bainite but is facilitated if martensite is produced. For transformation to martensite a very high cooling rate is required. The transformation behaviour of this steel has been computed from a database at TU Freiberg and it has been found that a cooling time of $t_{8/5} = 1.3$ s is necessary to obtain martensite. Problems in attaining the critical cooling rate have been encountered in the rolling of plates both on the mill at the Technical University at Freiberg and on the Corus 3-high mill at STC. In the first case, martensite is formed if plates no thicker than 7 mm are rolled; in such thin plates the thermal behaviour is unstable and the temperature difficult to control. At STC similar problems have arisen when using plates of too large a gauge without sufficiently rapid quenching. Achievement of this degree of temperature control thus seems very impractical.

- (c) Since Type 316 Stainless Steel is one of the slowest materials to recrystallise, the as-deformed structure in a single pass can be captured more readily by rapid quenching. However, even at the rates of $1 - 2 \text{ s}^{-1}$ typically used in the Gleeble machine and in laboratory rolling of a specimen at very high temperature, the time for significant recrystallisation may be less than the handling time required to re-engage in the rolls or quenching and limit the range of interpass times considered. Nevertheless, for moderately high temperatures and interpass times, the speed of recrystallisation is conducive to the detection and measurement of a range of levels of partial recrystallisation. If the temperature is too low, however, the yield stress of Type 316 stainless is high and there is a danger that the rolling load may break the mill even if moderate reductions are used.
- (d) The use of alternative materials which do not have the problems of CMnNb steel associated with transformation and recrystallise slowly was briefly considered. One such material is Fe30% Ni, which does not transform for the thermal conditions considered. Several variants have been produced experimentally but all recrystallise faster than Type 316 Stainless Steel. The selection/creation of such materials having analogous properties in relation to stress history but more transparent to experimental scrutiny must, however, remain on the agenda in future developments.

A1.4.4 Characterisation prior to processing

- (a) Recrystallisation kinetics of CMn and CMnNb steels

Experiments to test the suitability of Husain's method [A1.27] for investigating the material behaviour of CMnNb steel specimens deformed uniaxially at 1100°C and at a strain rate of 1 s^{-1} , with or without reversal, to a total strain of 0.4 have been referred to in Section A1.3.1.

Figure A1.3 shows the stress relaxation curves for both cases, showing the typical 'S' shape, with the reversal case (triangle symbol) having a higher stress than the non-reversal case (square symbol). Figure A1.4 shows also the well-known Bauschinger effect (a reduction in flow stress on the reversal direction) and a significant delay to the onset of dynamic recrystallisation. Figure A1.4 shows the (static) recrystallisation kinetics, which indicates that the reversal case takes around 45 seconds for full recrystallisation while the non-reversal case takes about 12 seconds.

The experiments confirm that stress relaxation is a suitable method for characterising the recrystallisation kinetics

The effects of main rolling variables (viz. composition, reduction, grain size, strain rate, temperature, interpass time/distance, etc.) on recrystallisation kinetics were investigated to assist in determining the rolling schedule on Freiberg's high speed rod rolling mill and laboratory plate rolling.

Metallurgical modelling for CMn, CMnNb and even medium carbon steels was used to express the activation energy, 50% and 95% recrystallisation fraction as a function of the above variables. The case considered was for an initially 20 mm diameter rod, rolled to oval 16 mm x 22 mm after the first pass F1 at the speed of 10 m/s (strain rate 100 1/s), the distance between the first pass F1 and the second pass F2 being 1500 mm (corresponding to an interpass time 0.15 s). The effects of temperature, initial grain size and strain were also investigated.

It was demonstrated that:

- (i) coarsening the initial grain size will significantly increase the recrystallisation time, which makes partial recrystallisation possible before entering to the next pass.
- (ii) increasing the temperature significantly reduces the recrystallisation time.
- (iii) decreasing the strain significantly increases the recrystallisation time. (cf. validation from the University of Oulu and from the assessment of the high speed trials).

Evolution of grain size of a medium carbon steel, soaked at 1200°C and quenched after 20 minutes was also investigated to assist the heat treatment process for the Freiberg high-speed rod rolling in achieving an initial grain size of 200 microns.

- (b) Metallographic examination of Type 316 stainless steel

Heavy etching of specimens in 'Newdip' reagent revealed a central core of different appearance to that at the quarter depth shown respectively in Figs. A1.63 and A1.64. A light etch showed that a second phase of delta ferrite was also present with some twinning of the grains (as seen in Fig. A1.65). There is also evidence of inclusions secondary phases delta ferrite and, in the interdendritic spacing, of the original as-cast macrostructure and segregation. This may incur difficulties in the initial definition of the grain size and in its measurement as it evolves. The mean grain-size as measured was low at 8.98 μm with a standard deviation of 0.58 μm . (The underlying results are given in Table A1.3.)

In view of the analysis above, it was decided to maintain use of Type 316 stainless steel for H-V rolling and bi-compression tests, but to seek material in more homogeneous initial condition and having a larger average grain size, by growing them if necessary. In view of its hardness, special care is needed not to exceed the operating limits of the mill.

A1.4.5 Material characterisation following deformation

(a) Strain reversal tests

Stress relaxation was used, for the CMnNb steel specimens, to obtain the volume fraction recrystallisation curves shown in Figs. A1.14–A1.16. From close inspection of these figures, the order in which tension and compression are applied makes little difference to the commencement and subsequent evolution of recrystallisation except for a slightly greater intermediate speed at 1000°C when compression is applied first. Implementation of the total strain without reversal (i.e. compression only) at the lower temperatures significantly delays the start of recrystallisation and may retard the initial speed.

(b) Multidirectional tests

The stress – time curves in Figs. A1.21 – A1.35 for Type 316 stainless steel have been discussed extensively in Section A1.3.2, the most notable results probably being (a) that a turn hastens early recrystallisation for large reductions but for small reductions retards it and (b) opposite behaviour to the results quoted in the literature in [A1.28] for CMn steels.

(c) H-V Rolling

Comparing the micrographs in Figs. A1.40, A1.42, A1.44, A1.53 – A1.56 and discussed in Section A1.3.3, for Type 316 Stainless steel after deformation, with those in Figs. A1.63–A1.65 discussed above it appears that the metallurgical structure is non-uniform at the outset and ambiguous after completion of the trials despite heat treatment prior to rolling to minimise heterogeneity. These observations are reflected in results of the EBSD tests shown in Figs. A1.57 and A1.58. Inverse Pole Figure maps were also used to examine the spread of orientations within grains, those having a small spread compared to the average for the grain being probably recrystallised. The maps for specimens with and without the extra reheat are shown in Figs. A1.66 and A1.67. The maps use colour to represent the orientation of solved patterns. Patterns that the software was unable to solve are coloured white. The figures unfortunately show that a large fraction of the patterns were not solved, indicative of poor pattern quality. The poor pattern quality, however, could be an indication of strain in the lattice, indicating that recrystallisation, but very little, has occurred. Because of the high fraction of unsolved patterns, it was considered unreliable to use the parameter (and associated maps) quantifying the grain orientation spread.

The above heterogeneity could have arisen because (a) the material is not homogenisable (b) the initial structure is not stable with respect to the thermomechanical schedules used (c) the techniques for material characterisation are not sufficiently sensitive for the steel qualities and schedules employed –or some combination of the above. The results however show a marginal tendency for intermediate turns to reduce recrystallisation but there is insufficient evidence that this is a primary effect.

(d) Forward -reverse rolling of plates

Six samples cut from the centre of the CMnNb plates rolled in the forward-reverse trials described in Section A1.3.3.2 were submitted for micro examination, hardness testing, and grain size measurement. After etching all the samples in 2% Nital reagent, examination revealed a wide range of microstructures between samples. Some samples also exhibited a variation in microstructure through the thickness of each individual sample. All samples were decarburised on both surfaces. In order to show the variations in microstructure a micrograph was taken of each sample at positions of 3 mm below the surface to avoid the

decarburisation and at the centre position. The surfaces were randomly marked as 1 and 2 for identification. Photomicrographs corresponding to trials 1, 4 7 and 8 are shown in Figs. A1.68 to A1.71.

The samples were then re-polished and etched in Japanese Etch to try and reveal the prior austenite grain size. This was only successful for 2 samples (viz. for trials 7 and 8) having martensitic type microstructures. Large differences in grain size were observed through the thickness of each sample, and it was concluded that grain size evaluation could not be undertaken to achieve meaningful results. Photomicrographs were again taken to show the variations observed, that for trial 7 is shown in Fig. A1.72.

Vickers hardness tests to BS EN ISO:6507-1:1998 were also carried out on each sample at positions of 3 mm below each surface and the centre. The results in Table A1.14 show the variations in observed microstructure. The effect of reversal on hardness was however inconclusive.

(e) High speed rolling

The experimental aspect of this work for CMn steel rods has been carried out by TU Freiberg and is discussed in Appendix 5. Grain sizes have been examined optically but were sometimes obscured by the presence of segregation. The results show that the increasing the roll speed has marginal, if any, effect on final austenite grain size.

A1.5 Formulation of constitutive model (Task WP4)

Various strategies available for constitutive modelling have been evaluated and the findings summarised in Table A1.2. From these findings, it has been decided to utilise the unified viscoplastic model, developed by Birmingham University in conjunction with Swinden Technology Centre [A1.20].

The model is based on physical mechanisms for dislocation dynamics, evolution of grain size, recovery, dynamic and static recrystallisation and their coupling with viscoplastic metal flow. The primary variables in the model are therefore the relative dislocation density $\bar{\rho}$, the instantaneous austenitic grain size d , the volume fraction recrystallised S , the isotropic hardening R , the elastic and plastic strain tensors ε_{ij}^e and ε_{ij}^p and the Cauchy stress σ_{ij}

Details of the model are contained in Ref. [A1.20]. The model is formulated via the following equations:

$$\dot{\varepsilon}_e^p = \left(\frac{\sigma_e - R - k}{K} \right)^n d^{-\gamma_4} \quad \dots (A1.17)$$

$$D_{ij}^p = 3S_{ij} / (2\sigma_e) \dot{\varepsilon}_e^p \quad \dots (A1.18)$$

$$\sigma_{ij}^J = G D_{ij}^e + 2\lambda D_{kk}^e \quad \dots (A1.19)$$

$$S = \{Q_0[\chi\bar{\rho} - \bar{\rho}_c(1-S)]\}(1-S)^{N_q} \quad \dots (A1.20)$$

$$\dot{\bar{\rho}} = (d/d_0)^{\gamma_d} (1 - \bar{\rho}) \left| \dot{\varepsilon}_e^p \right| - c_1 \bar{\rho}^{c_2} - (c_3 \bar{\rho} / (1-S)) \dot{S} \quad \dots (A1.21)$$

$$\dot{R} = B \dot{\bar{\rho}} \quad \dots (A1.22)$$

$$\dot{d} = \alpha_0 d^{-\gamma_0} - \alpha_2 \dot{S}^{\gamma_3} d^{\gamma_2} \quad \dots (A1.23)$$

where $\dot{\varepsilon}_p$ is the effective plastic strain rate and D_{ij}^p , D_{ij}^e are the rates of elastic and plastic deformation respectively, σ_{ij}^J is the Jaumann stress rate and χ is the variable controlling the onset of recrystallisation satisfying the differential equation.

$$\dot{\chi} = A_0 (1 - \chi) \bar{\rho} \quad \dots (A1.24)$$

$\bar{\rho}_c$ is the critical dislocation density, G and λ are the Lamé elastic constants, k is the yield stress and the other symbols are constants for the material modelled (currently Type 316 stainless steel). These are determined, for using a genetic algorithms program, from flow stress data obtained on the Gleeble machine at Swinden Technology Centre.

A1.6 Incorporation of constitutive models into FE models for application to complex loading paths (Task WP5)

The default material model used at Swinden Technology Centre in the ABAQUS FE code is based on an elasto-plastic formulation in which the flow stress changes instantaneously as strain/strain rate changes. This constitutive behaviour is entered either in tabular format or in combination of tables and analytical expressions for rate and/or temperature dependence of the flow stress. More complex constitutive models, e.g. viscoplastic, based on dislocation dynamics etc. are implemented by user defined material subroutines, viz. UMAT in the ABAQUS/Standard variant of the code (so called Implicit formulation) and VUMAT in the ABAQUS/Explicit version.

Flow stress in tabular format

There are two standard ways of incorporating these into the FE package, ABAQUS.

The first method is to generate flow stress-plastic strain tables from the constitutive equations for various constant temperatures and strain rates. This is limited to a uniform distribution of strain and strain rate because of ABAQUS' requirement of regularisation. The default tabular input is usually used in ABAQUS because of its simplicity. The stress-strain data at various strain rates and temperatures are entered following standard laboratory tests such as performed on a Gleeble thermomechanical simulator either in plane strain or uniaxial compression mode. The flow stress at a given strain and strain rate is then determined from interpolation within the extreme boundaries of the data. No extrapolation is made for any strain, strain rate or temperature outside the defined conditions. This therefore could result in underestimation of flow stress during hot rolling and could therefore result in discrepancies in load and torque predictions.

The second method is to write the user-defined subroutine UMAT, noting that special techniques of integration and stability are required for different types of constitutive models. A user defined material subroutine UMAT was available for the viscoplasticity constitutive model [A1.8] developed by Corus to represent the flow of steels at high working temperatures including flow softening associated with dynamic recrystallisation.. It has been shown that the ABAQUS-UMAT method gave more reasonable predictions of flow stress than the ABAQUS plasticity model. However, the numerical stability of the integration algorithm in the UMAT depends on deformation and loading condition, which converges quickly in some cases but, in general, converges slowly or even diverges, thus limiting its applicability to cases where detailed and accurate strain and stress distributions are required for the prediction of microstructural development.

The unified model based on dislocation dynamics developed at MEFOS [A1.29,A1.8] has also been implemented by means of a UMAT within ABAQUS/Standard v6.3 to allow simulation of kinematic hardening. Because of its relatively slow implementation in ABAQUS/Standard, it is recommended for simple sections only. Implementation within ABAQUS v6.3 using the *CREEP card was assessed as an easier option in particular, by using the elastic D matrix instead of the exact Jacobian, enabling reasonably large time steps to be made without significant loss in accuracy.

An alternative method considered of incorporating constitutive models into FE code was to use the commercial software Z-mat from NorthWest Numerics/Transvalor [A1.30]. This is a user-extendable material models' library that provides additional material modeling capabilities to ABAQUS. In particular, it allows the user to incorporate new materials within the standard Z-mat library and to make a permanent customised site library of UMAT routines which can then be combined with the hot rolling material models that have been developed. In addition, the Z-mat library provides an efficient integration scheme for the implicit solver.

Due to the availability of the trial version of the Z-mat software, work was carried out to assess the Z-mat features that could possibly be used in the current project. This was divided into two tasks viz.: -

1. Use of the standard Z-mat materials library by an FE rolling model.
2. Coding of a given material model and integration into the standard Z-mat library for use in ABAQUS simulation of rolling.

The viscoplasticity models in Z-mat were tested with the samples supplied by the software vendor and gave the expected results. An FE rolling model was also created with an attempt to use the Z-mat built-in materials models and to compare the CPU time with the standard tabular input. Work was not carried out, however, beyond the terms of reference of the trial.

An FE rolling model was subsequently created using the dislocation based viscoplasticity model formulated in A1.5 aimed at modelling the 2 pass bar rolling process described in Section A1.3.3.1. Data were therefore obtained from double hit tests, each of 0.43 strain, on Type 316 Stainless Steel specimens at 1000°C, with an interpass delay of 40 s. The stress-time and stress-strain curves are shown in Fig. A1.73. The data were then fed into the GA program to determine the coefficients for the model for this application; the constants obtained are given in Table A1.15.

A1.7 Model validation (Task WP6)

Finite element programs have been applied to several of the experimental tests and rolling trials to obtain predictions of the operating parameters and to ascertain the constitutive behaviour.

A1.7.1 Strain Reversal

The ability for modelling strain reversal was assessed using six approaches combining isotropic and kinematic models using tabular input of: (1) Isotropic hardening, (2) Linear kinematic hardening with a constant isotropic elastic domain, (3) Nonlinear kinematic with constant isotropic hardening, (4) Nonlinear kinematic and exponential isotropic hardening, (5) Nonlinear kinematic constant isotropic hardening, and (6) Nonlinear kinematic parametric input with nonlinear isotropic hardening. It was concluded that isotropic hardening should be re-formulated to accommodate the softening caused by recrystallisation for which a material user subroutine is required.

FE modelling of a simple geometry, cylinder and block, subjected to 'pulled-compressed', 'pulled-pulled' and bi-direction deformation conditions, was carried out using ABAQUS isotropic tabular input and the useful results were obtained to guide the strain reversal and bi-direction tests.

The results show that:

- (i) Strain reversal occurs.
- (ii) The equivalent plastic strain 'PEEQ', which is used to calculate the recrystallisation kinetics, does not reflect any change in history of strain components. For example, a cylinder subject to 'pulled-compressed' condition would have the same 'PEEQ' as that subjected to 'pulled-pulled' condition, thus resulting the same von Mises stress. However, to consider the Bauschinger effects induced by straining reversal, kinematic modelling methodology should be used.

A1.7.2 Multidirectional compression

FE Predictions of the 3 normal and 3 shear strain components have been obtained for the 40% reduction of the 10 mm cube at 1000°C, considered in Section A1.3.2 and are shown in Fig. A1.74(a-f). The normal strains are qualitatively as expected but the significant variability is exhibited in the shear strains.

A1.7.3 High speed rolling

The FE cooling model of the Freiberg two stands was created with a furnace discharge temperature 1250°C. The variation of temperature with time was examined to provide understanding of the temperature evolution during rolling. The FE model showed that the strain distribution over the cross section of rolled rod was non-uniform, highest in the centre (0.7) then the middle (0.55) then the surface (0.3). So if complete REX occurs in the centres, it may be only partially recrystallised at the surface. Predictions and more detailed information is given by TU Freiberg in Appendix 5.

A1.7.4 Forward reverse rolling

FE modelling of the plate rolling described in Section A1.3.3.2 was also carried out. Examples of predictions obtained of spread and equivalent plastic strain are shown in Figs. A1.75-78, namely for the 3 passes of Trial 1. Values of the predicted and experimental spread for these passes are contained in Table A1.16. For the 3 passes of Trial 4, predictions have been obtained using the in-house metallurgical model as shown below

Trial 4

Roll radius=155 mm

Speed=78 rpm

Rolling Temperature=1020°C, 3 passes with reduction from 30 mm to 25 mm, then to 22 mm and then to 19 mm.

Interpass time=10,12,12 s, respectively.

The following results were obtained where T50 and T95 are the times to 50% and 95% recrystallisation respectively.

Strain rate=9.6, 8.7, 10 1/s, respectively, for the three passes

Strain=0.21, 0.15, 0.17, respectively, for the three passes

T50=7.2, 4.5, 3.7, respectively, for the three passes

T95=60.2, 37.7, 31.2, respectively, for the three passes

Unrecrystallised fraction=0.42, 0.26, 0.52, respectively, for the three passes

Grain size=59, 47, 39 microns respectively, for the three passes

5%_Nb_ppt=24.1, 25.7, 30.6 s, respectively, for the three passes

Force=37.2, 31.3, 32.1 tonnes, respectively, for the three passes

Torque=10.2, 7.1, 7.2 kN m, respectively, for the three passes

The predicted loads and torques viz. 37.2,31.3 and 32.1 tonnes and 10.2,7.1 and 7.2 kN m respectively generally agree moderately well with the corresponding steady state measured values of 35, ~30 and 37.5 tonnes, and 8, 10 and 10 kN m (top roll)

A1.7.5 Stock rotation during rolling

A1.7.6 Dislocation based model

The dislocation based constitutive model formulated in Section A1.6 has been applied to some of the H-V rolling trials described in Section A1.3.3. In particular, it has been used to simulate the rolling, in two monotonic passes, of a 23 mm square bar of Type 316 Stainless Steel to a thickness of 15 mm. The thickness on exit from pass 1 is 18.5 mm and the time between the passes is 10 s. For both passes the rolls are of diameter 154 mm and rotate at a speed of 50 rev/min. A uniform temperature of 1100°C has been assumed throughout.

The model was implemented in ABAQUS v6.4 and predicted distributions of stress and strain components including Mises stress and PEEQ strain, dislocation density, fraction recrystallised and recrystallised grain size. Examples of these distributions viz. for PEEQ strain, Mises stress, dislocation density and grain size in Pass1 when the stock is fully engaged by the rolls are shown for the 2 passes in Fig. A1.79(a-d). These figures indicate, not surprisingly, essential uniformity in the distributions of equivalent strain, grain size and dislocation density, the values of strain being slightly greater towards the surface than at the centreline and higher than the value employed previously in analytical models of

recrystallisation. The distribution of Mises stress is not uniform, reaching its maximum near mid radial position in the arc of contact.

Similar predictions for Pass 2 are contained in Fig. A1.80 except that Fig A1.80(d) shows the fraction recrystallised. The strain distribution is more non-uniform than in the first pass. The predicted distributions of dislocation density distribution and fraction recrystallised are slightly non-uniform, greatest values occurring towards the surface, the latter demonstrating, in particular, that the material is essentially unrecrystallised as indicated in the results of the H-V rolling trials.

A1.8 Summary

The various complex loading paths assigned in the Technical Annex have been implemented by means of a series of experimental tests and laboratory rolling trials of the allocated steel qualities to assess their effect on operating parameters and evolution of microstructure specifically grain size and recrystallisation, with outcomes as listed below:

1. In strain reversal tests on CMnNb steel specimens, subjected, on the Gleeble, alternately to tensile and compressive strains of equal magnitude (0.2) at temperatures of 1000°C, 1050°C and 1100°C.
 - Notable Bauschinger effects were observed.
 - The magnitudes of these effects, and subsequent REX behaviour are independent of the order of application of the strains and marginally dependent on temperature.
 - The recrystallisation kinetics are delayed if a strain of the same total magnitude is applied without reversal, but, for the lower temperature, are faster overall.
2. In single and multiple compression tests of various magnitude on Type 316 stainless steel, conducted isothermally on the Gleeble at temperatures of 900°C, 1000°C and 1100°C, stress relaxation tests indicate that:
 - The time for recrystallisation in a single hit decreases more significantly with strain than with temperature.
 - Recrystallisation behaviour is marginally dependent on the order of the hits.
 - Multiple hits of the same total magnitude as a single hit give overall slower recrystallisation and may inhibit recrystallisation entirely if, individually they are moderately large (~20%).
 - An interpass turn of 90° hastens early recrystallisation when one of the reductions is large (~40%) but significantly retards it when both hits are moderate (~20%).
3. In H-H/H-V rolling of Type 316 stainless steel bars rolled on the STC Cavendish mill to successive reductions of 20% at temperatures between 950°C and 1100°C, indications exist that interpass turns and intermediate low reheats retard microstructural evolution but are not strongly supported by the experimental evidence.

- In bar rolled at $\sim 1100^{\circ}\text{C}$ and given a turn of 90° in the transverse plane in 10 s interpass, lower recrystallisation is observed at the surface than without the turn but this is not supported by lower loads in the second pass.
 - In bars rolled first at 1040°C , reheated to 950°C before the second pass, larger grain sizes with an interpass turn are deduced than can be attributed to retarded grain refinement.
 - No differences in loads in the second pass are observed between bars given or not given an intermediate reheat to 1040°C and grain sizes differ little from those before rolling suggesting that any recrystallisation for either case is insignificant.
4. In the rolling of CMnNb steel specimens on the STC plate mill, loads with or without reversal are very similar suggesting that reversal has no effect on REX and load or that any differences are masked by thermal effects.
 5. In the high speed rolling of CMn rods, at TU Freiberg, grain sizes may marginally be increased with increasing roll speeds but effects can be masked by segregation.
 6. Difficulties in drawing sound conclusions on the influence of strain path on structural evolution may be related to problems associated with characterisation of the materials allocated for investigation viz.
 - In CMnNb steels, the austenite grain size may be masked by phases such as bainite but is more detectable if martensite is present; however, the formation of the latter requires very rapid cooling rates which are achievable, e.g. for plates, for thicknesses less than 7 mm in which temperatures are very unstable and difficult to control.
 - In Type 316 stainless steel, recrystallisation, although slower than most other steel qualities, may, if temperatures or interpass times are large, may give rise to partial but not full recrystallisation, which can be discriminated in trials with or without interpass turns.
 - In Type 316 stainless steel material inhomogeneity occurs in the original samples, associated with inclusions, delta ferrite etc. which may not be significantly reduced even with heat treatment and may give rise to unpredictable variations in grain size in the rolled samples and make it difficult to establish, either by optical metallography or EBSD, whether or not recrystallisation has occurred.
 7. Constitutive models, both analytical and numerical, implementable in the ABAQUS FE code, have been applied in the simulation of some of the above tests with reasonable agreement on the operating parameters. Their scope for further development/application for the efficient and accurate prediction of microstructural evolution has been reviewed. The dislocation based viscoplastic model, developed in conjunction with Birmingham University has accordingly been selected and applied to the H-H/H-V bar rolling trial using constitutive data obtained from results of delayed double hit tests on the Gleeble under similar conditions. Predictions of equivalent strain, dislocation density, fraction recrystallised and recrystallised grain size have been obtained for both passes and confirm that, in the second pass, the material is essentially unrecrystallised.

A1.9 Conclusions and recommendations

A range of models has been developed not only providing good agreement with experimental measurements of load and deformation for rolling cases considered, but breaking new ground in generating reasonable predictions of the evolution of microstructural quantities such as dislocation density and fraction recrystallised, and having the potential to analyse this evolution in regions of stock undergoing complex loading paths as in section rolling.

To realise this potential:

1. More experience should be gained in applying the model to, and acquiring constitutive data for, single and multipass operations involving complex loading histories and abrupt changes in local strain paths.
2. More importantly, much attention needs to be paid to refining and advancing the techniques for material characterisation, for selecting appropriate materials for characterisation and for stringently controlling and monitoring the heating/cooling and rolling operations used not only in the laboratory but also in an industrial context.
3. Differences between the results from experimental mechanical tests and rolling trials need to be investigated and explained.

A1.10 References

- A1.1 Orowan, E: Proc. IME, 1943, Vol. 150(4), 140.
- A1.2 Ekelund, S: Steel, (1933), Vol. 93, 8.
- A1.3 ABAQUS Theory Manual v6.1, ABAQUS, Pawtucket, USA.
- A1.4 Krausz, A S and Krausz, K (eds.): 'Unified Constitutive Laws of Plastic Deformation', Academic Press, 1996.
- A1.5 Sherby, O D and Miller, A K: 'Combining Phenomenology and Physics in Describing the High Temperature Mechanical behaviour of Crystalline Solids', *ASME J. Eng. Mat. Technol.*, 1979, Vol. 101, pp387-394.
- A1.6 Sellars, C .M and McG. Tegart, W J: 'Hot Workability', *Int. Met. Rev.*, 1972, Vol. 17, pp1-24.
- A1.7 Garofalo, F: 'An Empirical Relation defining the Stress Dependence of Minimum Creep Rate in Metals', *Trans. American Inst. Mining & Met. Engineers*, 1963, Vol. 227, pp351-356.
- A1.8 Ingham, P M: 'The Effect of Strain Reversal and Strain-Time Path on Constitutive Relationships for Metal Rolling/Forming Processes', Final Report ECSC Project, 7210.EC/811, 2000.
- A1.9 Miller, A K: 'An Inelastic Constitutive Model for Monotonic, Cyclic, and Creep Deformation: Part I- Equations Development and Analytical Procedures; Part II-

- Application to Type 304 Stainless Steel', *ASME J. Eng. Mat. Technol.*, 1976, Vol. 98, pp97-113.
- A1.10 Dunne, F P E, Nannah, M M and Zhou, M: 'Anisothermal Large Deformation Constitutive Equations and their Application to Modelling Titanium Alloys in Forging', *Phil. Mag. A*, 1997, Vol. 75, no. 3, pp587 –610.
- A1.11 Farrugia, D C J: 'Finite Element Modelling of Deformation and Microstructural Evolution in Multipass Rolling', ECSC Project EUR 18398 EN, Final Report, p195 (formulae of S. Jaiswal and R. Barbosa).
- A1.12 Barbosa, R and Sellars, C M: 'Recrystallisation '92', p461.
- A1.13 Jaiswal, S et al: 'Modelling of Microstructure in Rod Rolling of Alloy and Stainless Steels', ECSC Project7210.EB/807 (D3.1/89) 1992.
- A1.14 Humphreys, F J and Hatherly, M: 'Recrystallisation and Related Annealing Phenomena', Pergamon, 1995.
- A1.15 Kolmogorov, A N: *Izv. Akad.Nauk,USSR-Ser_Matnat* 1(3),355, (1937). (also Johnson, W A and Mehl, R F: *Trans. Metall. Soc. A.I.M.E* 135,416,1939 and Avrami, M: *J. Chem. Phys*, 7,1103,(1939).).
- A1.16 Abramowitz, M and Stegun, I: 'Handbook of Mathematical Functions', Dover Publications, Inc, NY.
- A1.17 ABAQUS User Manual v6.1, ABAQUS., Pawtucket, USA.
- A1.18 Zhou, M, Ingham, P M and Farrugia, D C J: 'A User Defined Material Model in ABAQUS to Represent the Viscoplastic Deformation and Work Softening during Hot Rolling', SL/M/R/S474/1/98/D, British Steel plc, Swinden Technology Centre, July 1998.
- A1.19 Lemaitre, J and Chaboche, J L: 'Mechanics of Solid Materials', CUP,1994, pp253-345.
- A1.20 Liu, Y, Lin, J, Farrugia, D C J and Zhou, M: 'Modelling of Microstructural Evolution in Multipass Rolling', *School of Manufacturing and Mechanical Engineering, University of Birmingham, to be published.
- A1.21 Chaboche, J L and Rousselier, G: 'On the plastic and viscoplastic constitutive equations - Part I: Rules developed with internal concept, Part II: Application of internal variable concepts to the 316 stainless steel', *ASME J. of Pressure Vessel Technology*, 1983, Vol. 105, pp153-164.
- A1.22 Sherby, O D and Miller, A K: 'Combining Phenomenology and Physics in Describing the High Temperature Mechanical behaviour of Crystalline Solids', *ASME J. Eng. Mat. Technol.*, 1979, Vol. 101, pp387-394.
- A1.23 Estrin, Y: 'Dislocation-Density-Related Constitutive Modelling', in *Unified Constitutive Laws of Plastic Deformation*, Academic Press, London, 1996, pp69-105.

- A 1.24 Kopp, R: 'The Realistic Simulation of Metal Forming Process Chains', Steel Research, 1998, Vol. 69, pp121-127.
- A1.25 Kendall, M and Stuart, A: 'The Advanced Theory of Statistics', publ Griffin 1962.
- A1.26 'Computational Stochastic Mechanics in a Meta-Computing Perspective'. ed. J Marczyk, publ. CIMNE, 1997.
- A1.27 Husain, Z and Howe, A A: 'Use of Thermomechanical Simulation for the Validation of Microstructure Evolution Models', SL/PH/R/S0340/1/92/A, British Steel Technical, Swinden Laboratories, September 1992.
- A1.28 Dzubinsky, M: Comparison of recrystallisation kinetics established by stress relaxation, double hit, optical metallography methods and using orientation maps obtained by SEM/EBSD.
- A1.29 Wang, X, Siwecki, T and Engberg, G: 'A physical model for prediction of microstructure evolution during thermomechanical processing', Materials Science Forum. 2003, Vol. 426-432, pp3801-3806.
- A1.30 'User Manual: Z-mat Version 8.2', Transvalor/ENSMP and Northwest Numerics, Inc., Fall 2000. <http://www.nwnumerics.com>.

Table A1.1(a-c): Constants for Type 316 Stainless Steel in Equations (A1.8 –A1.11)
for recrystallisation

(a) Zener- Hollomon and static recrystallisation

Constants	316(a)	316(b)
Q_{def}	460E3	456E3
A	$1.45E-17$	4E-15
k	1.4	1
l	0.38	0.38
m	1.33	1.33
n	3.6	3.6
Q_{rex}	525E3	475e3

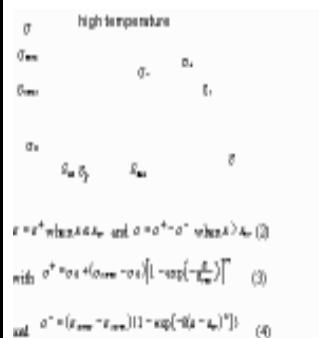
(b) Critical strain for dynamic/static recrystallisation

Constants	316(a)
A_c	2.5E-3
l_c	0.75
m_c	0.13

(c) Static recrystallised grain size

Constants	316(a)	316(b)
A_r	66	470
m_r	0.5	0.3
n_r	0.75	1
l_r	0.1	0.1

Table A1.2: Options for constitutive modelling

Code	Standard	Explicit	Comments	Current state	Priority
Creep	Does not require Jacobian, routines. Can implement in ABAQUS range* of models e.g. current Birmingham model. Implicit time integration Can use USDFL to define changes in internal state variables such grain size	Not available	*Excluding kinematic hardening. Can inc. dislocation models	Readily constructible (hence base method). Implement with microstructural internal SDV's (DynRx,GS etc)	
UMAT	Needs calculation of Jacobian and stress tensor, integration time may be prohibitively long but will do kinematic hardening	Not available	complex mathematical formulation. And coding. Can include dislocation models	(1) Version with previous viscoplastic model with rex and isotropic hardening prohibitively slow but (2) prototype code based on Birmingham model recently developed. At present only for simple sections. (3) Model with kinematic hardening should be available later	
VUMAT	Not relevant	Available	Only requires stress tensor at end of increment and internal/dissipation energy radial return technique	(1) Model available (analytical) or readily constructible(3) Can model kinematic hardening or isotropic hardening and probably viscoplastic flow by update of flow rule..Useful for complex sections such as beams	
Application to complex sections	No	Yes			
Application to simple sections, rod, bars, strip, 2D	Yes	Yes			
Application to 3D plate	Possible but needs many elements in explicit code	Yes			
Currently developed	UMAT isotropic hardening, REX	Vumat with analytical model 			

Possible models	Creep implementation with microstructural internal state variables (DynRx, GS, etc.) UMAT visco monotonic + same as above Umat kinematic hardening + visco + as above VUMAT + micro + Iso VUMAT + micro + kinematic Dislocation base models (Birmingham, Mefos, or Estrin) in Creep, UMAT and VUMAT				

Table A1.3: Evolution and distribution of recrystallisation in specimen 4R1F287 rolled in 2 passes without turn at 1100°C

Position:		Quarter depth		Central position			
Sample subcode	Reheat time (s)	Grain size (microns)	Standard Devn. (microns)	Grain size (microns)	Standard Devn. (microns)	REX (%)	Standard Devn. (%)
A	0	61.9	14.1	20.8	2.6	61.7	7.1
B	5	66.5	9.9	31.9	1.1	59.1	13.8
C	60	36.7	3.5	31.5	0.9	-	-

(Note: blank entry means not possible to measure meaningful value)

Table A1.4(a and b): Evolution and distribution of recrystallisation in specimen 4R1F50 rolled in 2 passes with turn at 1100°C

(a) Location 1

Position:		Quarter depth		Central position			
Sample subcode	Reheat time (s)	Grain size (microns)	Standard Devn. (microns)	Grain size (microns)	Standard Devn. (microns)	REX (%)	Standard Devn. (%)
A	0	57.2	3.9	26.7	3	43.1	16.6
B	5	34.5	2.0	25.1	1.6	—	—
C	60	37.6	2.9	33.6	3.2	—	—

(b) Location 2

Position:		Quarter depth		Central position			
Sample Subcode	Reheat time (s)	Grain size (microns)	Standard Devn. (microns)	Grain size (microns)	Standard Devn. (microns)	REX (%)	Standard Devn. (%)
A	0	65.1	4.4	28	3.9	70.9	4.5
B	5	81.5	8	22.8	0.5	—	—
C	60	37.7	1.4	31.9	1.1	—	—

Table A1.5: Grain size statistics for homogenised type 316 stainless steel bars

Sample	Average ($\times 10^{-3}$ mm)	Standard deviation	95% CL	% RA	ASTM Grain Size
4R1F 64B	91.4	10.6	13.2	14.4	3.9
4R1F 71B	101.7	7.3	9.1	8.9	3.3
4R1F 284B	102.6	13.4	16.7	16.3	3.3
4R1F 514B	91.0	10.8	13.5	14.8	3.6

Table A1.6: Rolling trials carried out with and without interpass reheat to 1040°C and with and without interpass turns

QA No.		Process history/(results)				
Code	Subcode	Preheating/ rolling temperature	Rolling: Pass 1	Interpass details (micrographs, heating profiles)	Turning operation	Rolling: Pass2 (Loads, Torques)
4R1F71		1040°C	23.1 to 18 mm	Q, Reheat to 1040°C and Hold 2 min, Q (Fig. A1.5, Fig. A1.7, Fig. A1.6)		
4R1F64A	1	1040°C	23.1 to 18 mm	Q, Reheat to 1040°C and Hold 60 s, Q, Reheat to 950°C (Fig. A1.8, Fig. A1.9)	No Turn	18 to 14 mm (Fig. A1.10(a and b))
4R1F284A	1	1040°C	23.1 to 18 mm	Q, Reheat to 950°C	No Turn	18 to 14 mm (Fig. A1.11(a and b))
4R1F284A	2	1040°C	23.1 to 18 mm	Q, Reheat to 950°C	90° Trans-Vert. Plane	25.5 to 20.4 mm (Fig. A1.12(a and b))
4R1F64A	2	1040°C	23.1 to 18 mm	Q, Reheat to 1040°C and Hold 60 s, Q, Reheat to 950°C	48° Trans-Horiz. Plane	18 to 14 mm
4R1F514		1040°C	22.6 to 14.2 mm			

Table A1.7: Average and standard deviation of grain size measurements made in specimen 4R1F284A_2, rolled in 2 passes, 1st pass at 1040°C from 23.1 mm to 18 mm, interpass quench, reheat, 90° turn, 2nd pass at 950°C from 25.5 mm to 20.4 mm

Position:		Quarter depth	
Sample subcode	Reheat time (s)	Av. grain size (microns)	Standard devn. (microns)
1	0	78.2	8.8
A	10	76.0	7.0
B	60	82.5	6.2

Table A1.8: Average and standard deviation of grain size measurements made in specimen 4R1F64A1, rolled in 2 passes from 23.1 mm to 18 mm, 1st pass at 1040°C, interpass quench, reheat to 1040°C, held for 60 s, reheat to 950°C

Position:		Quarter depth	
Sample subcode	Reheat time (s)	Av. grain size (microns)	Standard devn. (microns)
1	0	55.0	4.8
A	10	55.7	2.4
B	60	55.8	2.6

Table A1.9: Average and standard deviation of grain size measurements made in specimen 4R1F64A2, rolled in 2 passes, 1st pass at 1040°C from 23.1 mm to 18 mm, interpass quench, reheat, 45° turn in horizontal plane, 2nd pass at 950°C from to 18 mm

Position		Quarter depth	
Sample subcode	Reheat time (s)	Av. grain size (microns)	Standard deviation (microns)
(1)	10	52.6	5.5
A	20	56.6	3.9
B	60	56.5	6.6

Table A1.10: Measurements of randomly selected grains in specimen 4R1F284A1_1 rolled in 2 passes from 23.1 mm to 14.0 mm, 1st pass at 1040°C, between passes quench and reheat to 950°C, no turn

1 mm from edge (Equiaxed) Ferret max. (µm)	6 mm from edge (Equiaxed) Ferret max. (µm)	12 mm from edge (Mixed) Ferret max. (µm)	Ferret min. (µm)
110	174	476	204
247	192	466	217
155	273	314	156
154	259	258	127
95	169	270	98
209	191	508	159
146	179	314	145
99	124	241	58
95	281	430	192
89	171	249	93
92	157	142	56
114	303	214	118
119	240	240	98
150	191	216	70
158	228	184	56
112	139	160	77
130	176	41	14
113	150	24	13
132	90	21	14
104	140	21	11
120	213	39	19
117	193	38	14
95	205	30	14
91	167	28	12
70	199	26	14
114	311	24	11
200	177	17	13
85	211	18	12

84	208	20	12
59	268	16	10
81	185	36	10
101	189	21	9
53	211	174	39
59	156	17	9
57	381	23	16
57	230	31	23
109	223	40	20
97	188	46	17
68	171	29	23
85	178	36	12
135	161	18	11
137	121	27	18
87	310	51	29
109	157	46	13
88	217	33	12
138	202	31	17
65	192	28	20
145	182	26	12
90	412	27	15
110	126	15	7
94	138	56	28
	220		
	176		
	196		
	139		
	188		
	176		
	214		
	266		
	157		
	108		
	133		
	175		
	146		
	133		
	135		
	108		

Table A1.11: Chemical composition of CMnNb steel

C	Si	Mn	P	S	Cr	Mo	Ni	Al	Cu	Nb	N	Sn	Ti	V
0.15	0.37	1.40	0.016	0.009	0.017	0.002	0.02	0.033	0.007	0.03	0.006	0.002	0.001	0.003

Table A1.12: Chemical composition of type 316 L stainless steel

Steel	C	Si	Mn	P	S	Cr	Ni	Mo	N	Ti	V
316L	0.02	0.44	1.73	0.03	0.01	17.25	11.2	2.2	0.03	0.001	-

Table A1.13: Quantitative metallograph results for type 316 stainless steel specimen
(Micro JobK03M00208)

Traverse no.	Grain size (μm)	% 2nd phase
1	9.14	4.39
2	9.39	5.63
3	8.01	6.32
4	9.44	5.98
5	8.92	5.11
Average	8.98	5.48
Standard deviation	0.58	0.76

Table A1.14: Hardness results HV30 for specimens in forward – reverse rolling trials

Sample	Surface 1	Centre	Surface 2
T2Z8 T1	295	197	191
	307	197	194
	263	192	198
Average	288	195	194
T2Z8 T4	187	189	187
	189	186	186
	184	186	187
Average	187	187	187
T2Z8 T5	191	189	216
	194	191	216
	194	196	213
Average	193	192	215
T2Z8 T6	404	323	375
	385	307	373
	383	307	377
Average	391	312	375
T2Z8 T7	328	287	406
	329	295	396
	329	278	415
Average	329	287	406
T2Z8 T8	196	191	198
	195	189	195
	198	192	194
Average	196	191	196

Table A1.15: Parameters employed in application of constitutive model to 2 pass rolling of Type 316 Stainless Steel

PARAMETER	VALUE
E	$0.976586e5$
ν	0.34
k	19.164
d_0	40.5
p_c	0.28823
Q_0	9.9946×10^{-3}
N_q	1.2827
A_0	3413.8
K	15.126
N	7.8292
B	143.27
c_1	17.967
c_2	2.6931
c_3	14.539
α_1	3.2009
α_2	1368.9
γ_1	173.76
γ_2	0.13113
γ_3	7.7011
γ_4	2.2089
γ_d	7.1906

Table A1.16: Table A1.2: Plate reversal rolling measurements and FE predictions of spread for passes 1-3 of Trial 1

Pass	Experimental spread	FEM pass spread	Cumulative FEM spread	hm (mm)	L	hm/L
1	0.4	1.82	121.82	27.40	28.16	0.97
2	1.9	0.8	122.56	23.55	19.5	1.20
3	3	1.17	123.68	20.9	20.66	1.01

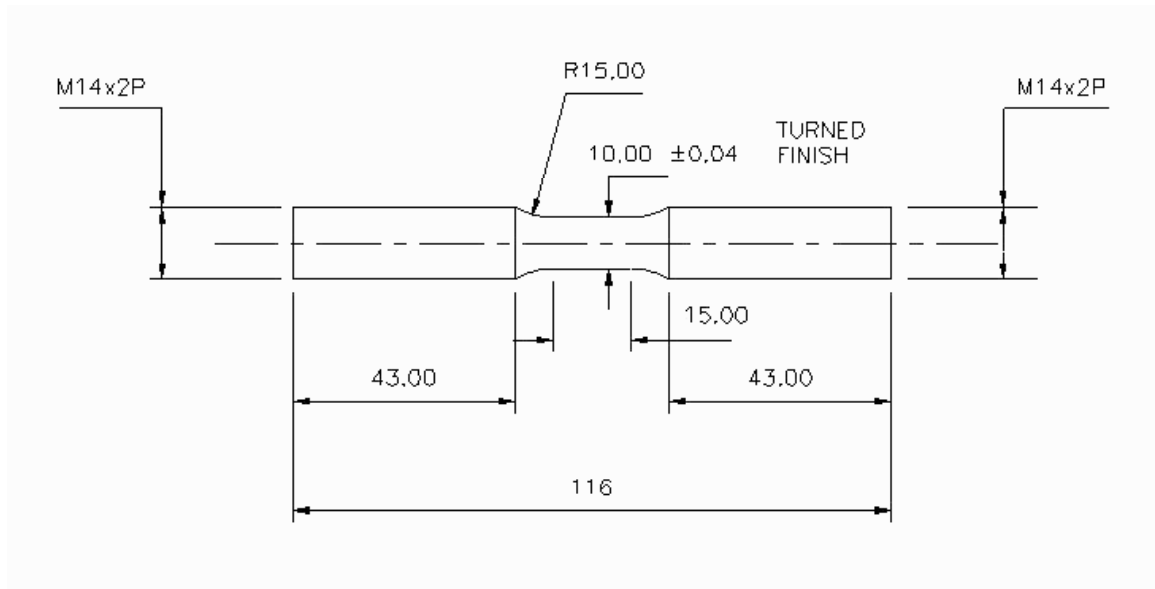


Fig. A1.1: Fatigue specimen for strain reversal tests



Fig. A1.2: Strain reversal test specimens subjected to compression and tension

Stress Relaxation

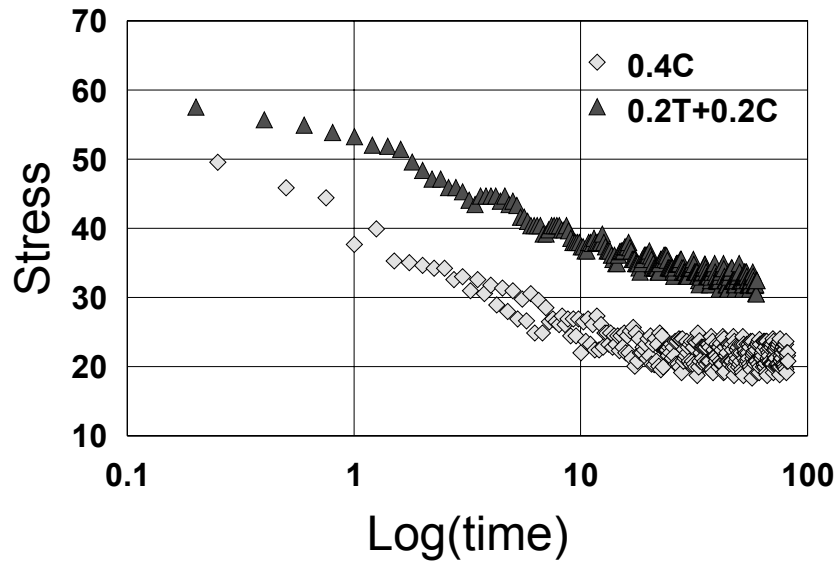


Fig. A1.3: Stress relaxation tests on CMn-Nb steels

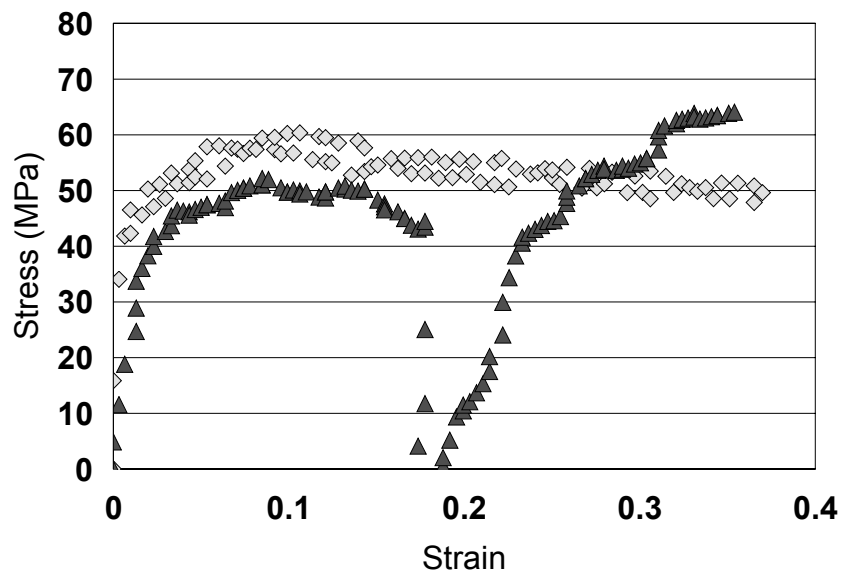


Fig. A1.4: Stress-strain curves (CMn-Nb steel)

REX Kinetics

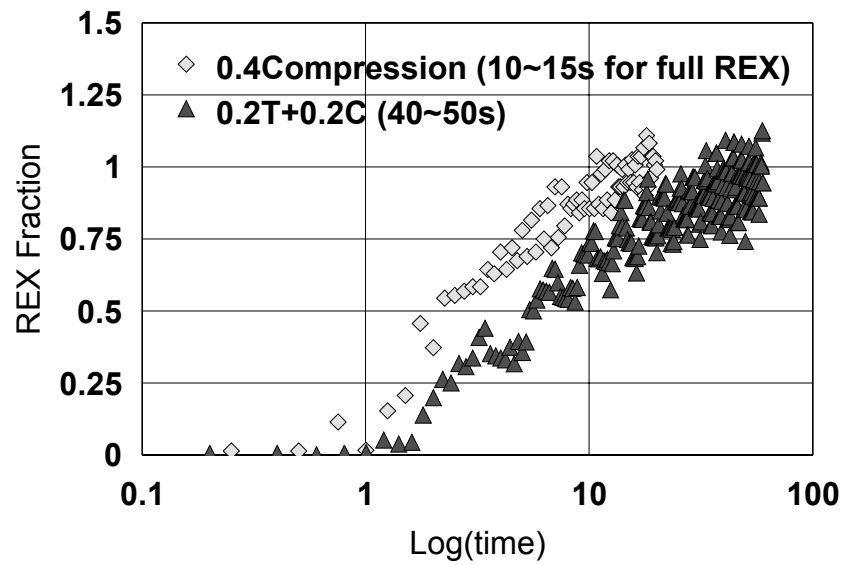


Fig. A1.5: Recrystallisation kinetics (CMn-Nb steel)

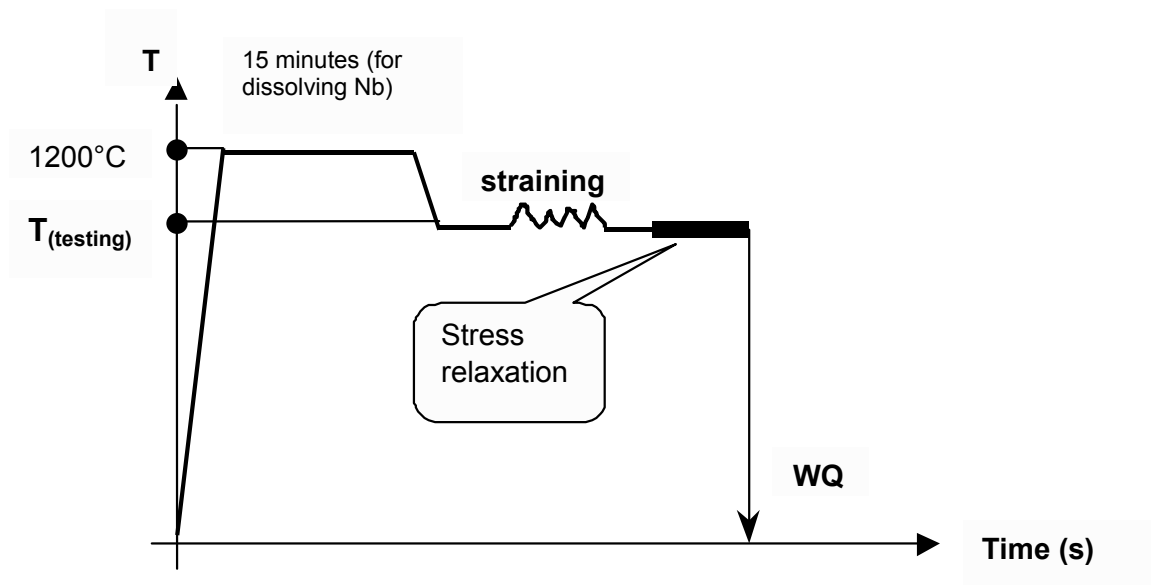


Fig. A1.6: Testing schedule for strain reversal tests

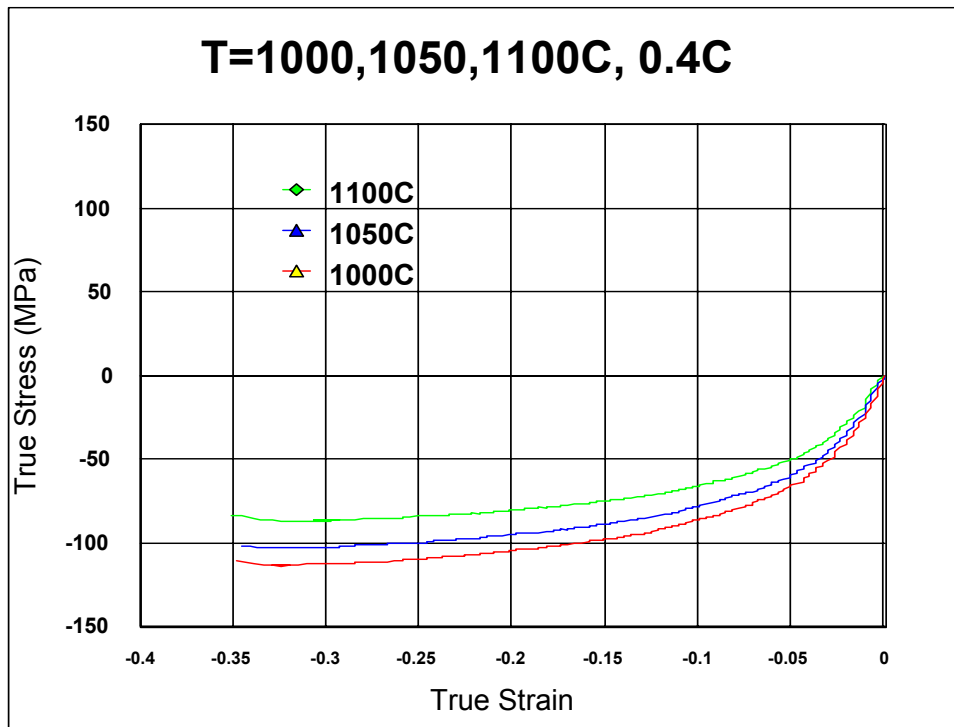


Fig. A1.7: Compression tests at different temperatures

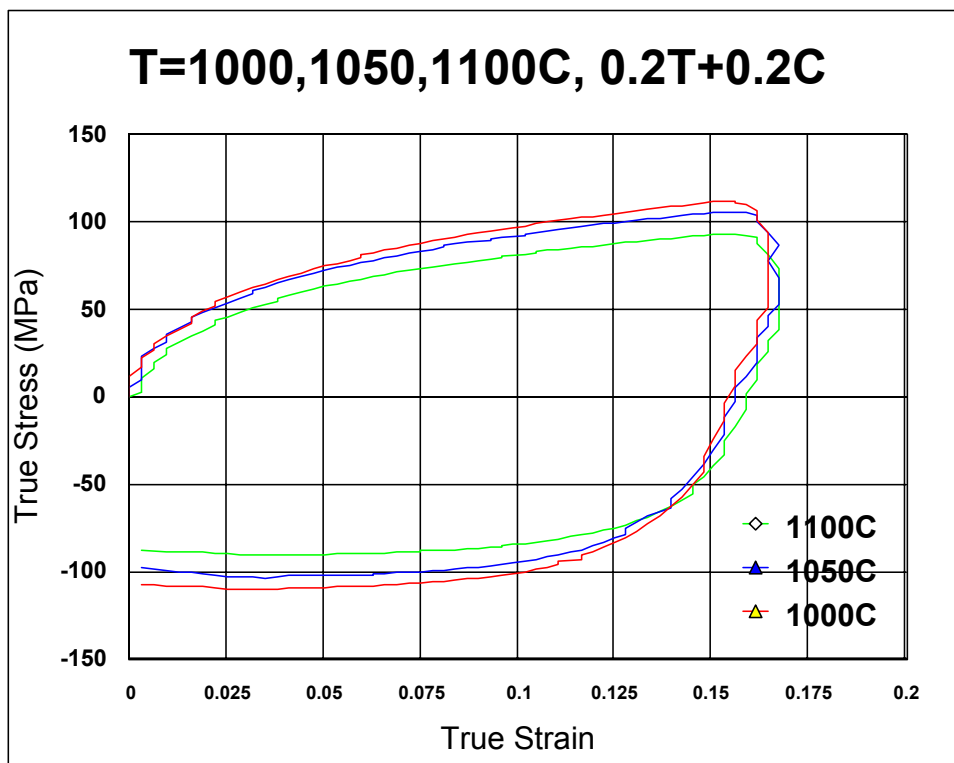


Fig. A1.8: Strain reversal tests 0.2T+0.2C

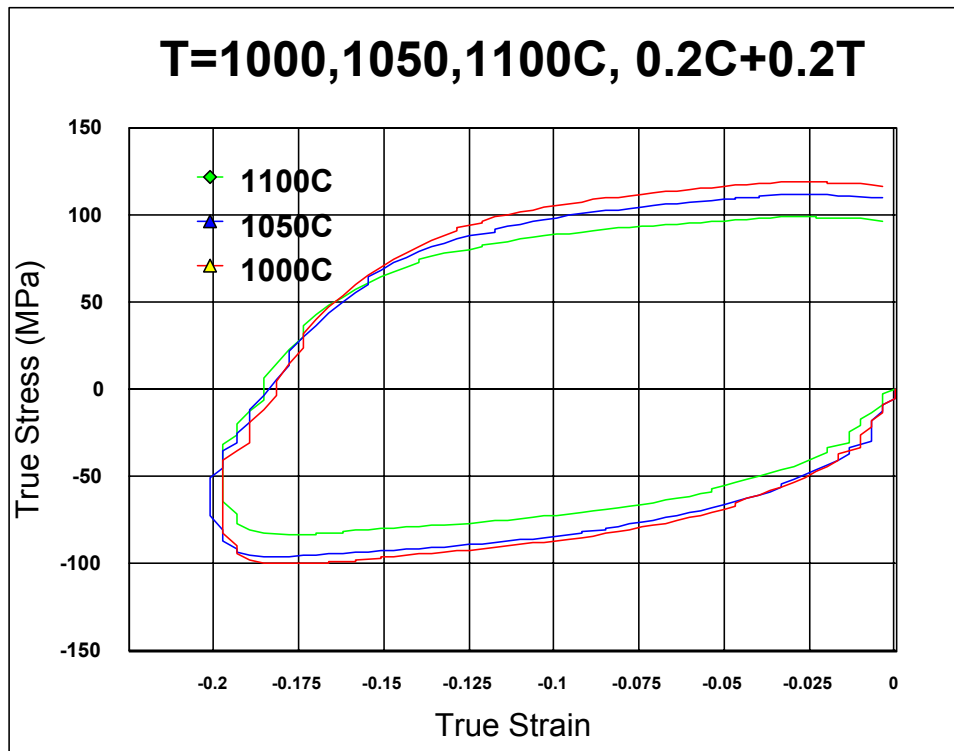


Fig. A1.9: Strain reversal tests 0.2C+0.2T

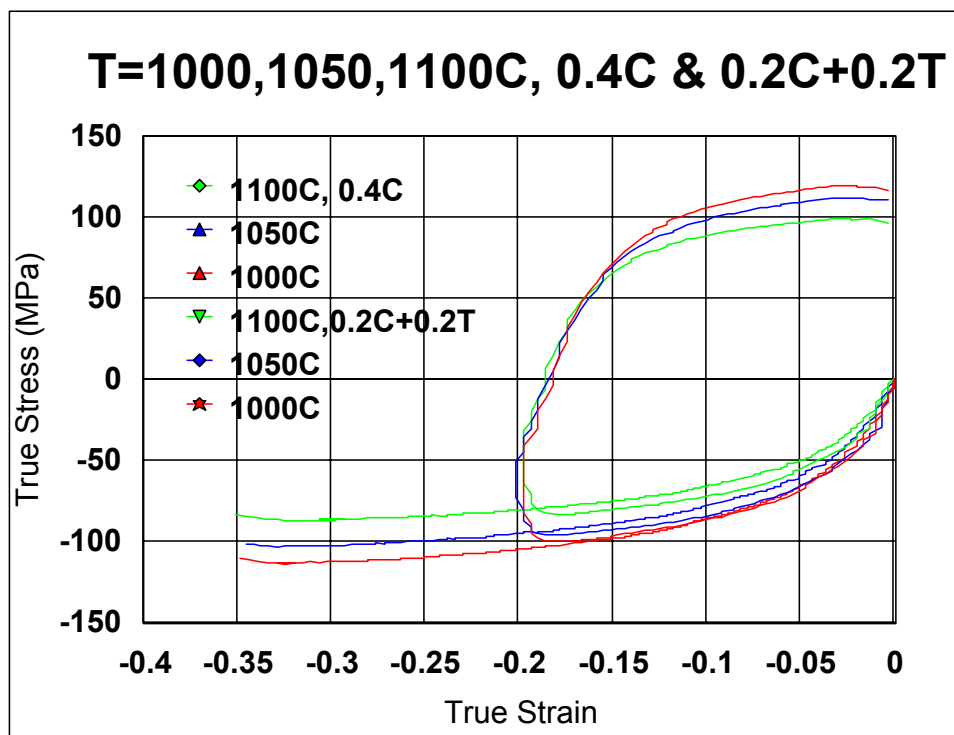


Fig. A1.10: Comparison of compression and reversal behaviour (1)

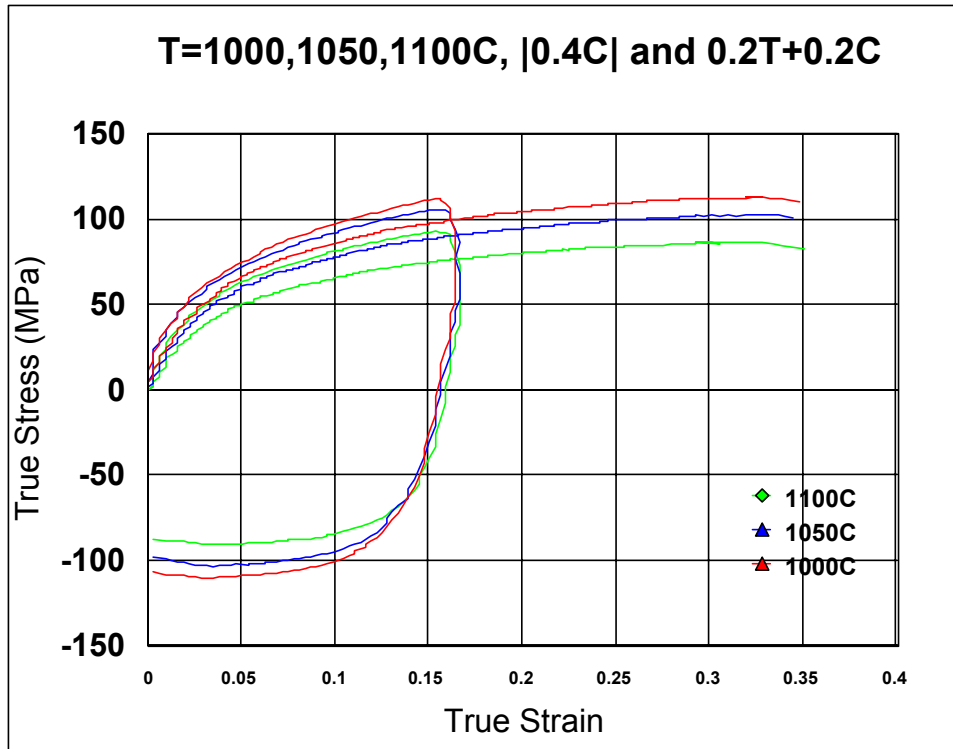


Fig. A1.11: Comparison of compression and reversal behaviour (2)

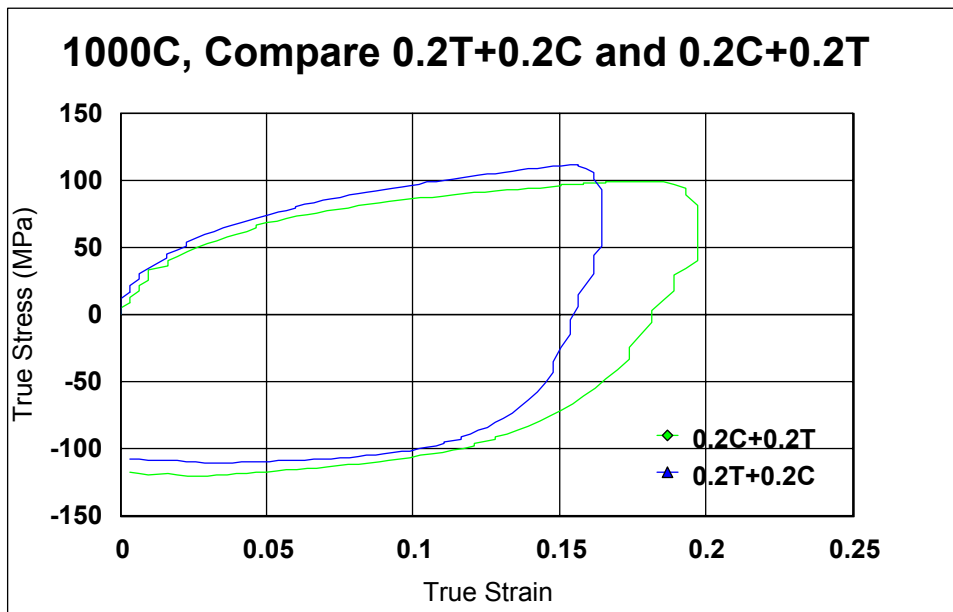


Fig. A1.12: Comparison of reversal behaviour: 0.2T+0.2C and 0.2C+0.2T

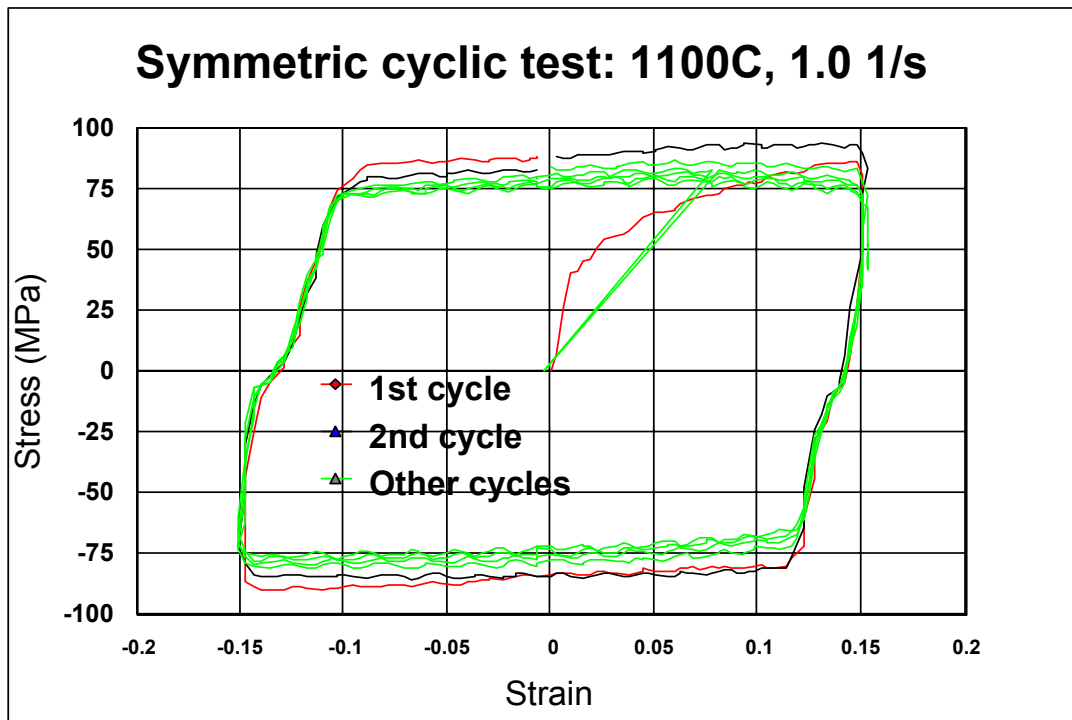


Fig. A1.13: Symmetric cyclic test

Comparison: 1000C, 1 1/s, Recrystallisation kinetics

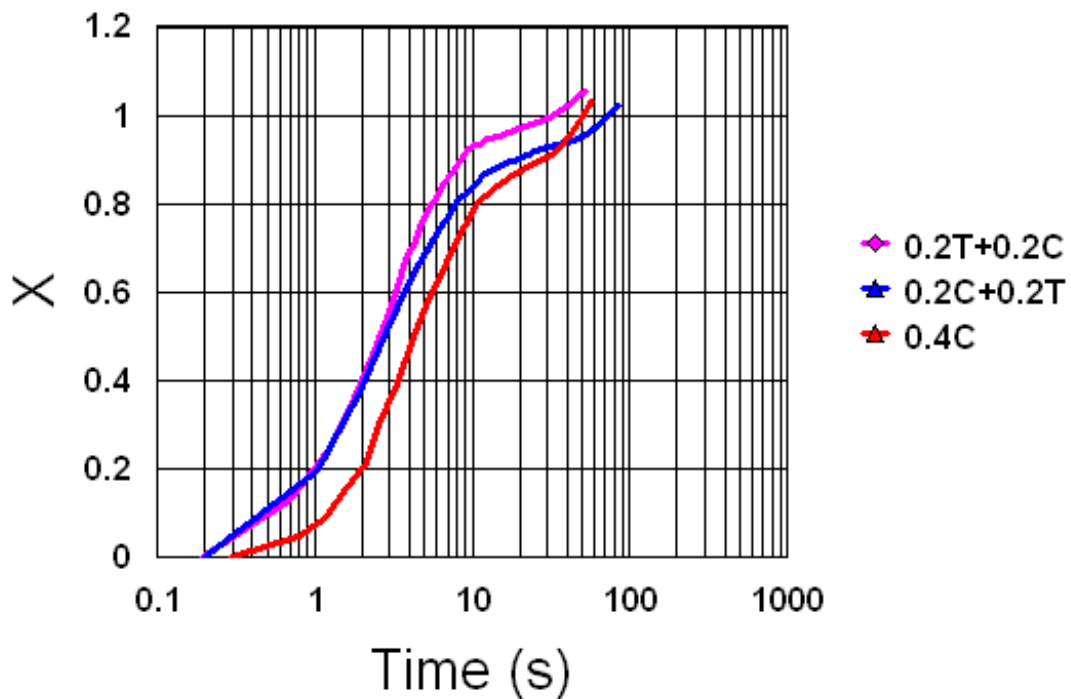


Fig. A1.14: Comparison of recrystallisation kinetics at 1000°C

Comparison: 1050C, 1 1/s, Recrystallisation kinetics

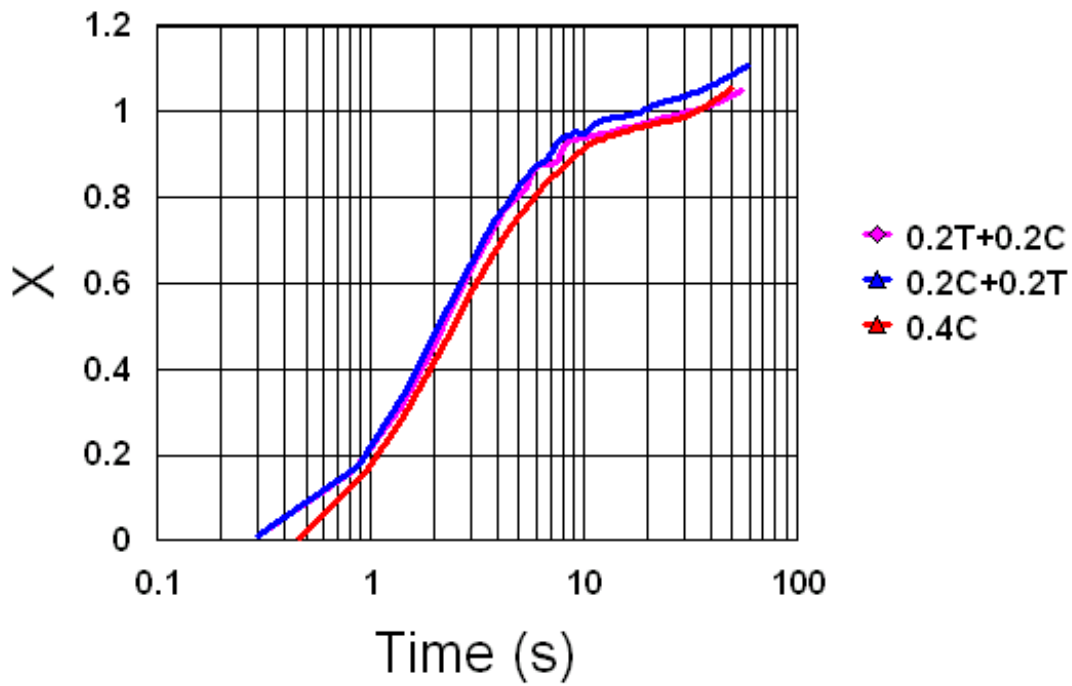


Fig. A1.15: Comparison of recrystallisation kinetics at 1050°C

Comparison: 1100C, 1 1/s, Recrystallisation kinetics

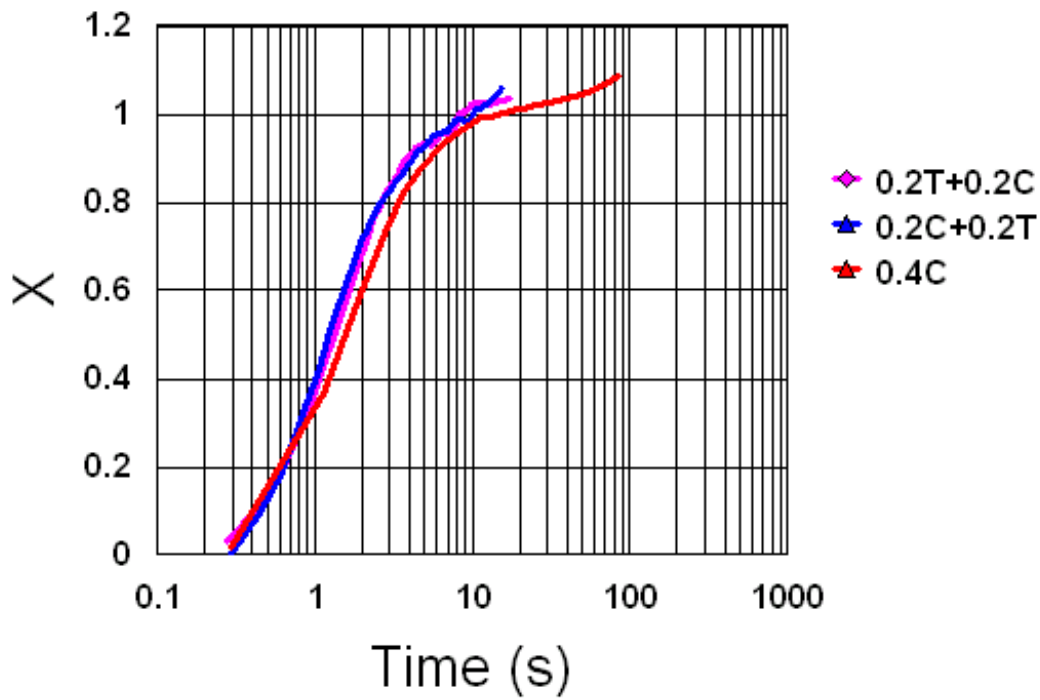
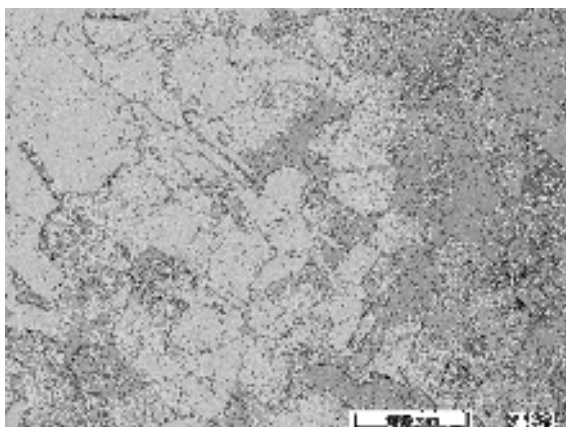
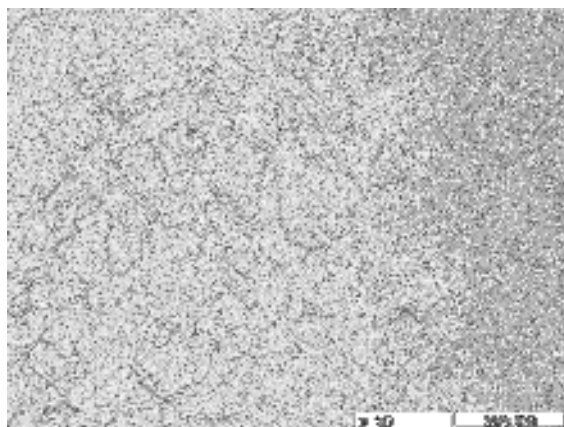


Fig. A1.16: Comparison of recrystallisation kinetics at 1100°C

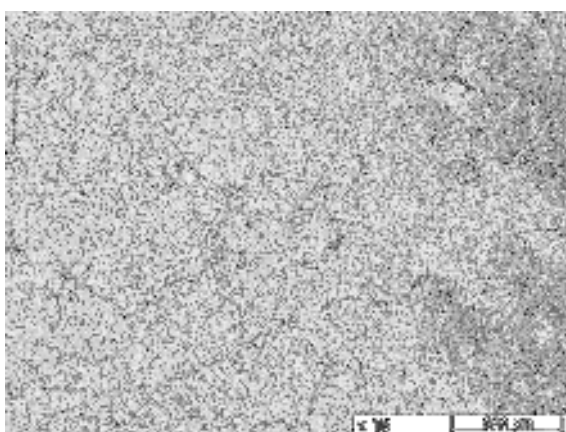


(a) 1200C/900C/0.3 Strain/Q

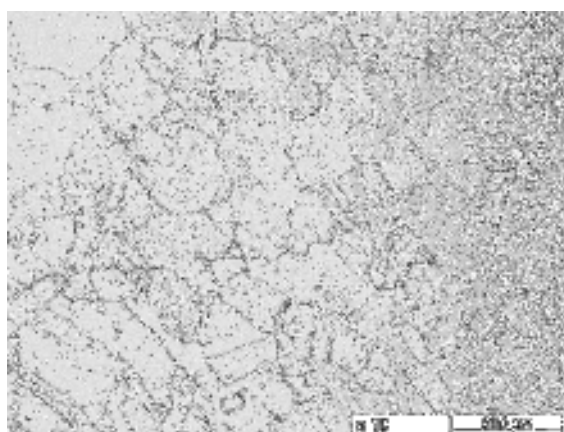


(b) Repeat of (a)

Fig. A1.19(a and b): Metallography of specimen quenched after the first compression

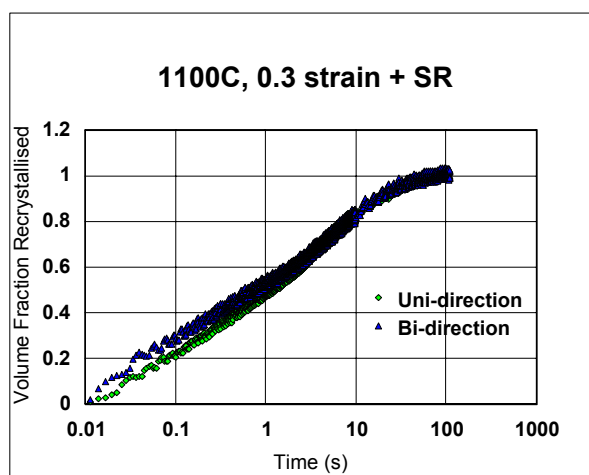


(a) 1200C/900C/0.3 Strain/Q

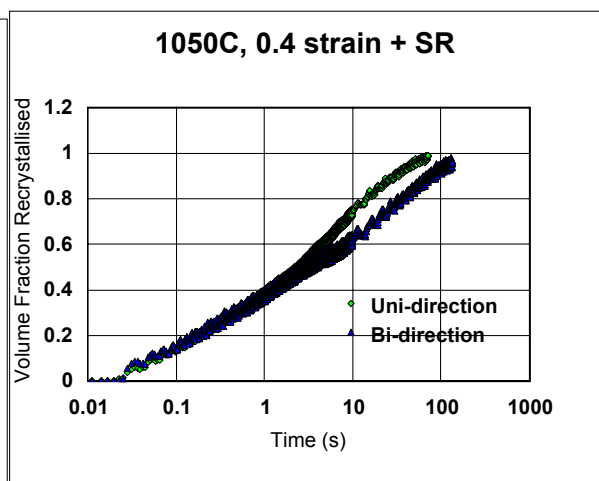


(b) Repeat of (a)

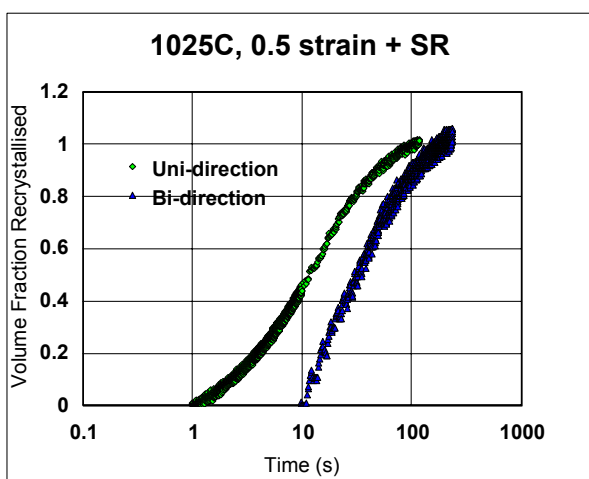
Fig. A1.20(a and b): Metallography of specimen in fig. A1.31 after 90 degree turn, reheated and quenched



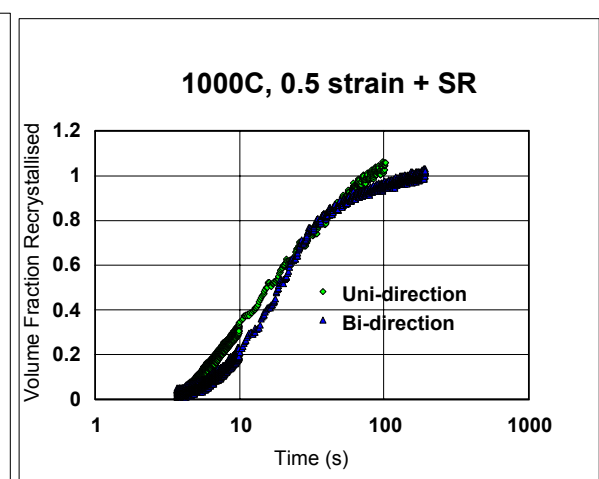
(a) 1100C



(b) 1050C



(c) 1025C



(d) 1000C

Fig. A1.21(a-d): Comparison of recrystallisation kinetics in H-V compression

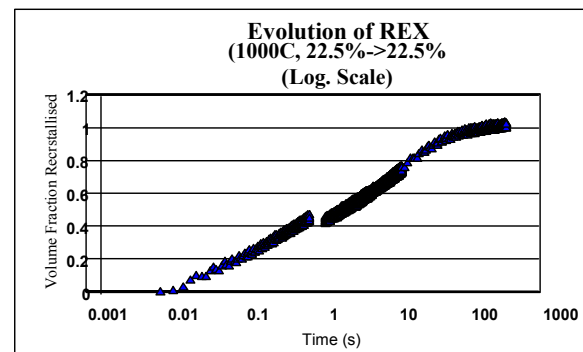
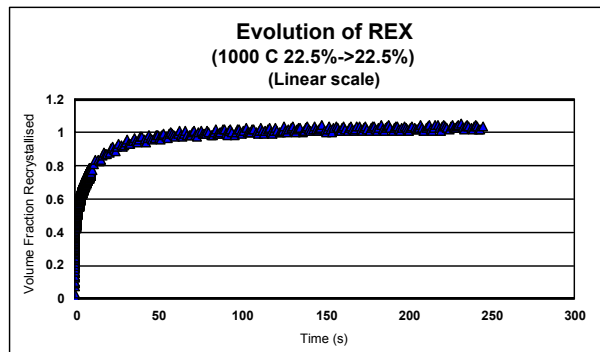
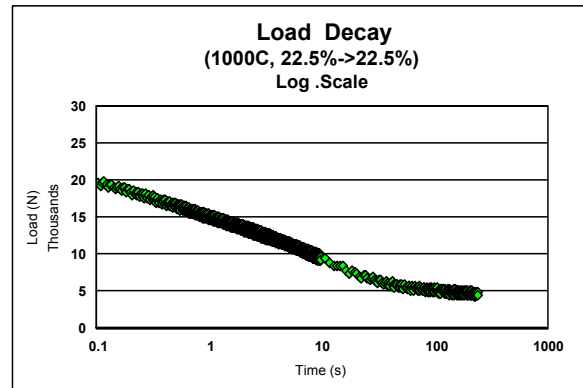
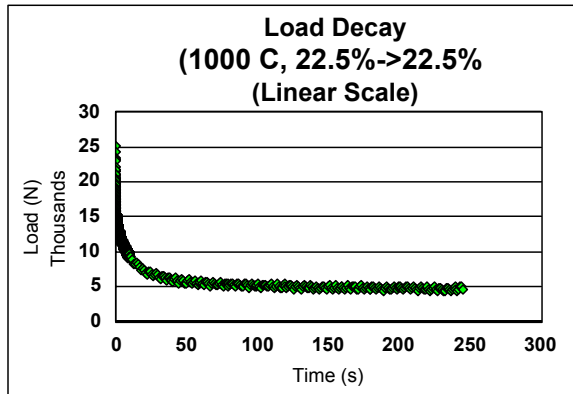
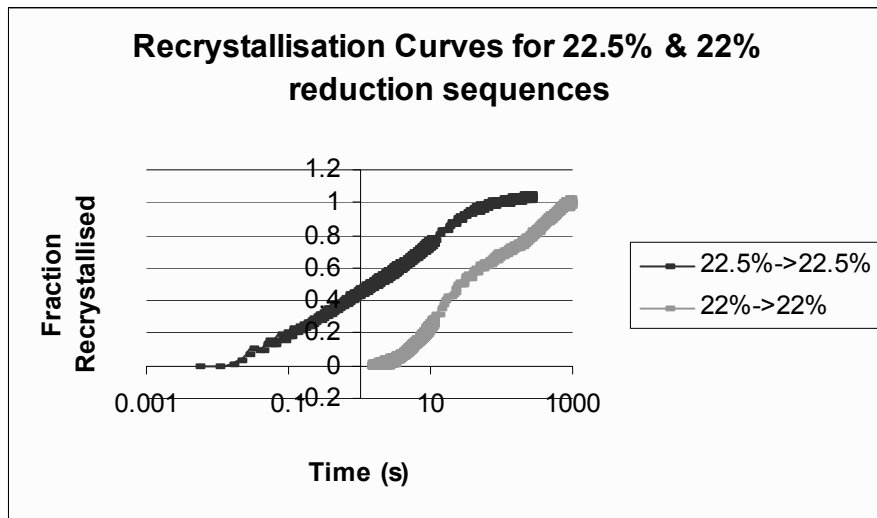
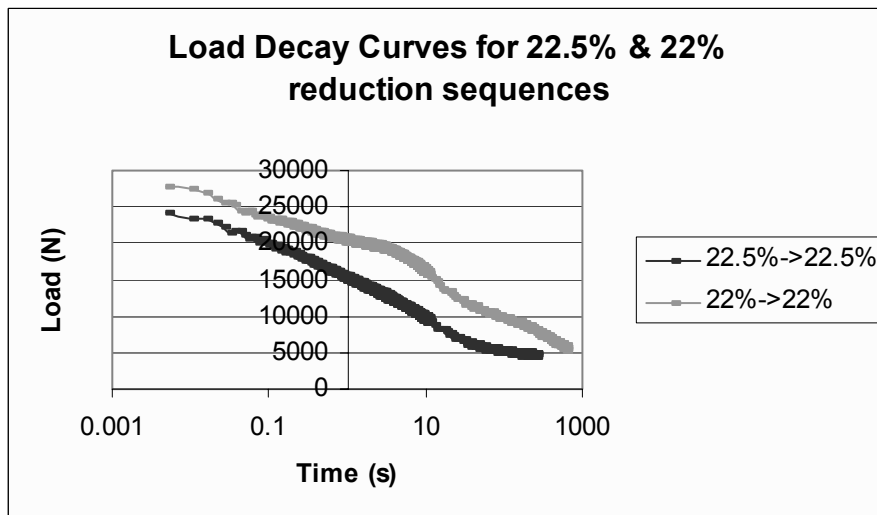


Fig. A1.22: Evolution of load and recrystallisation, obtained by stress relaxation, in specimen having undergone repeated compression of 22.5% at 1000°C.

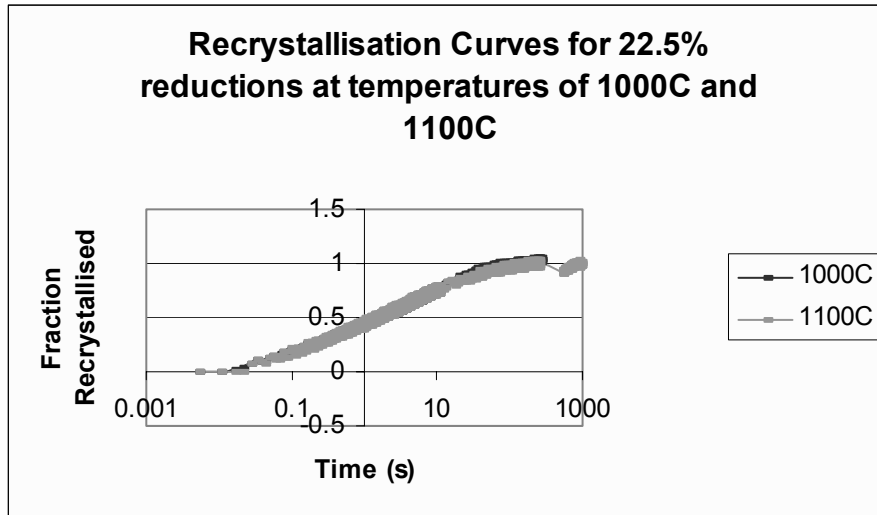


(a) Evolution of recrystallisation

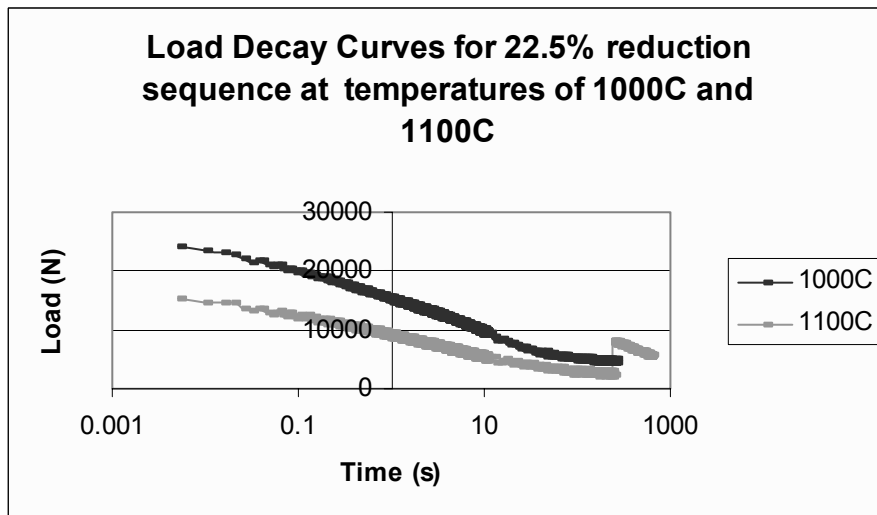


(b) Load decay

Fig. A1.23(a and b): Curves showing evolution of recrystallisation and decay of load for 2 reduction sequences: (1) two consecutive 22.5% compressions, (2) two consecutive 22% compressions



(a) Recrystallisation evolution



(b) Load decay

Fig. A1.24(a and b): Curves showing evolution of recrystallisation and decay of load for repeated compression of 22.5% at temperatures of 1000°C and 1100°C

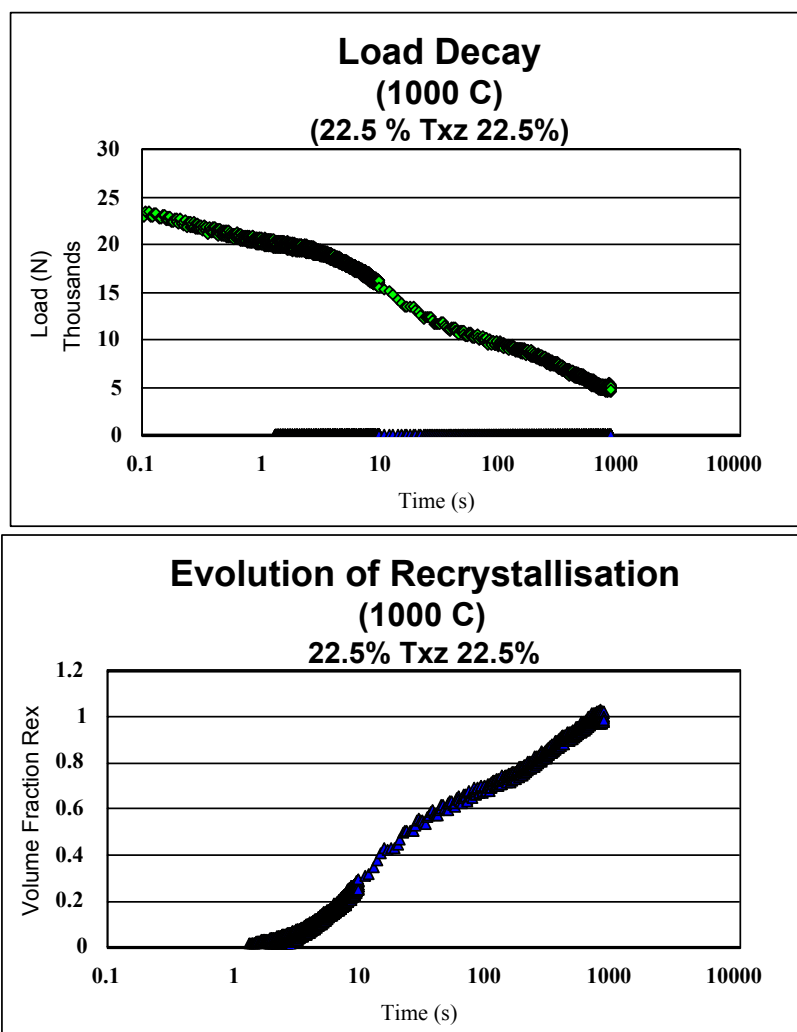


Fig. A1.25: Evolution of load and recrystallisation for repeated 22.5% compression at 1000°C with 90 degree interpass turn

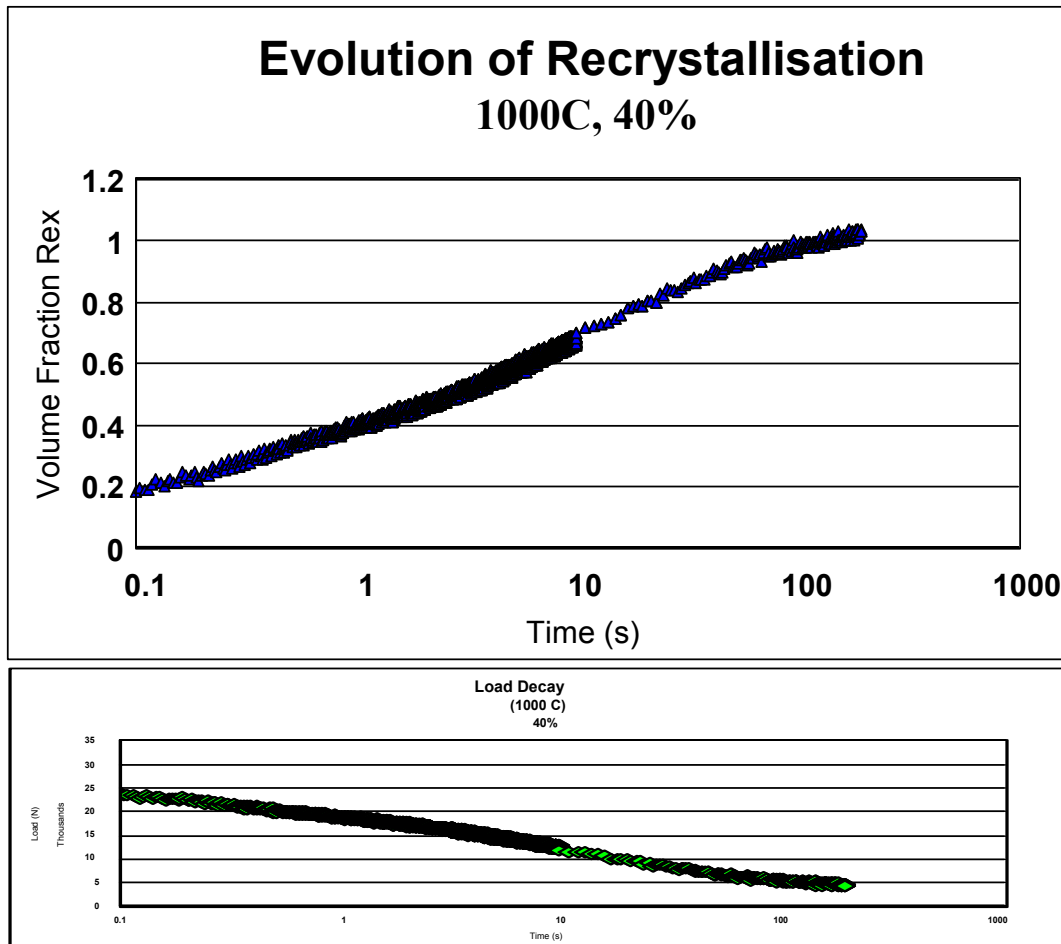


Fig. A1.26: Evolution of load and recrystallisation for single compression of 40% at 1000°C

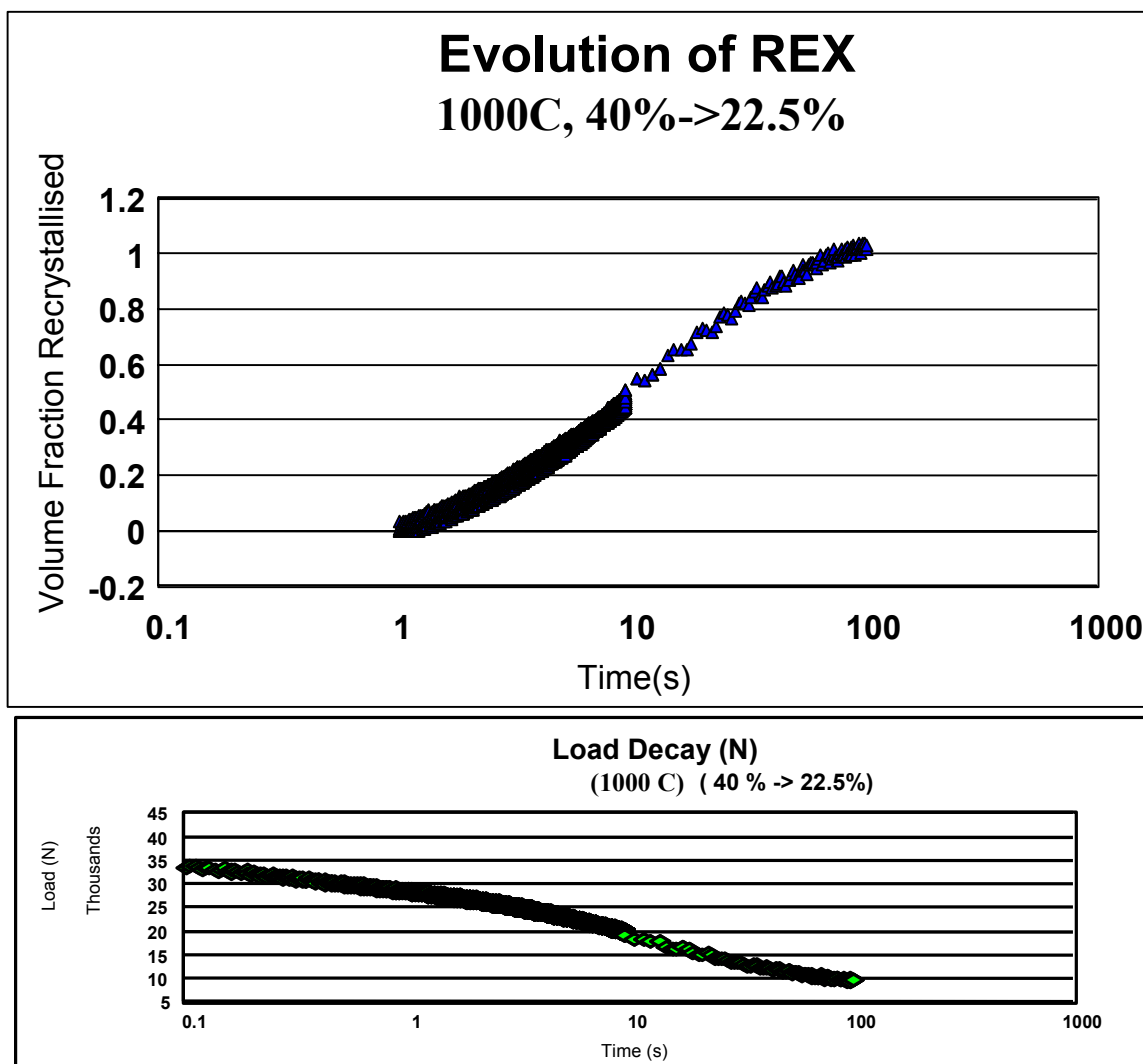


Fig. A1.27: Evolution of load and recrystallisation for 40% followed by 22.5% compression at 1000°C

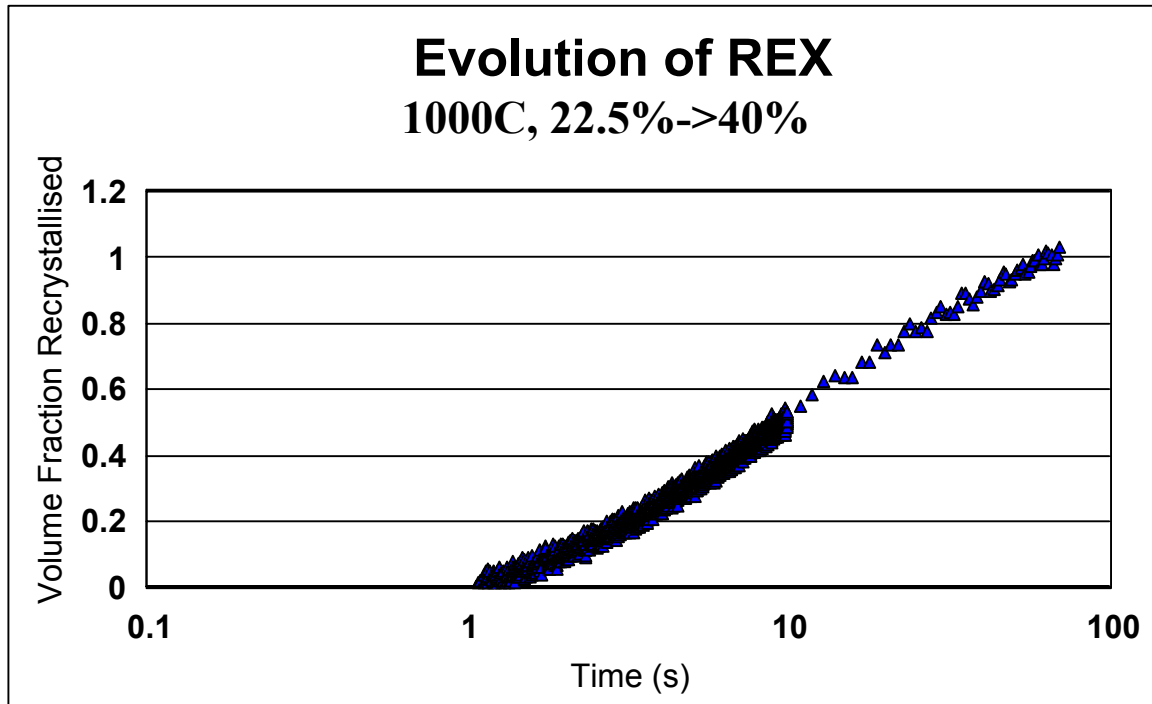


Fig. A1.28: Evolution of load and recrystallisation for 22.5% followed by 40% compression at 1000°C

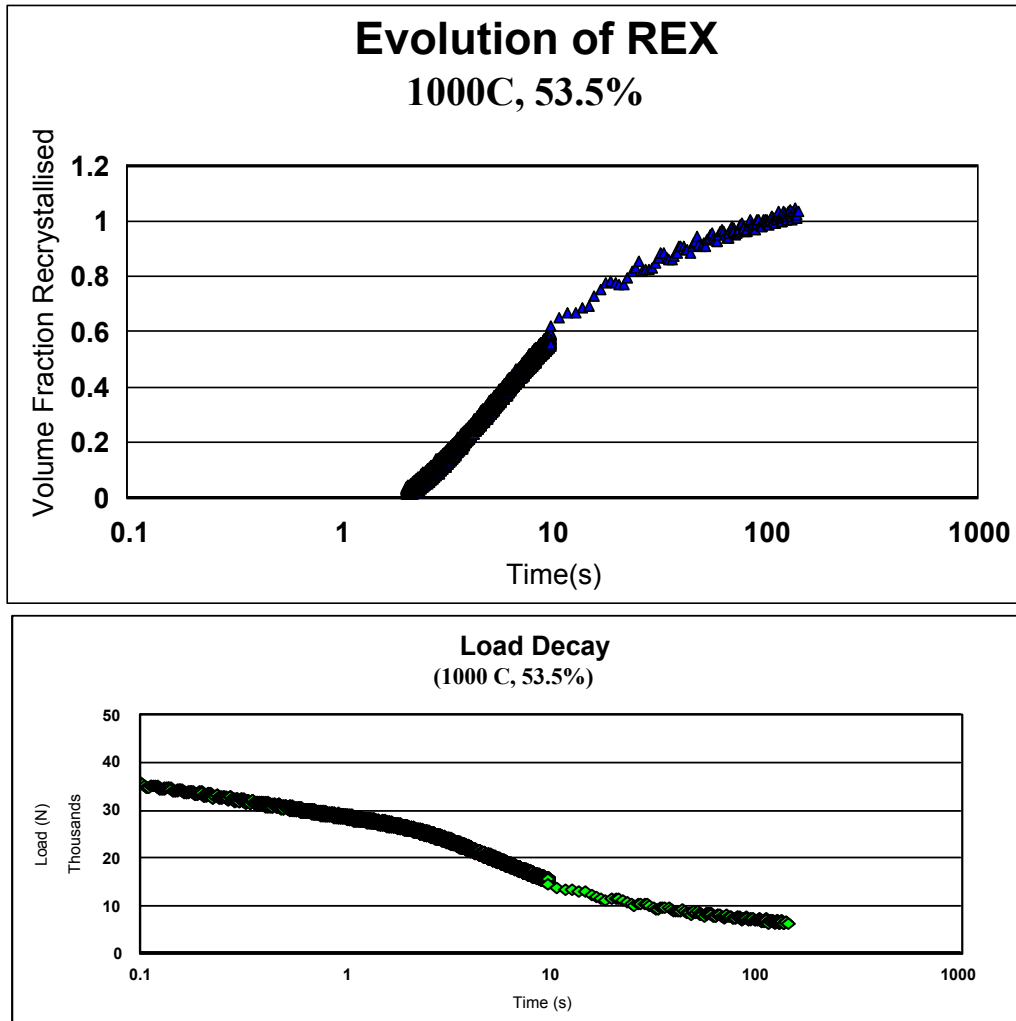


Fig. A1.29: Evolution of load and recrystallisation for single compression of 53.5% at 1000°C

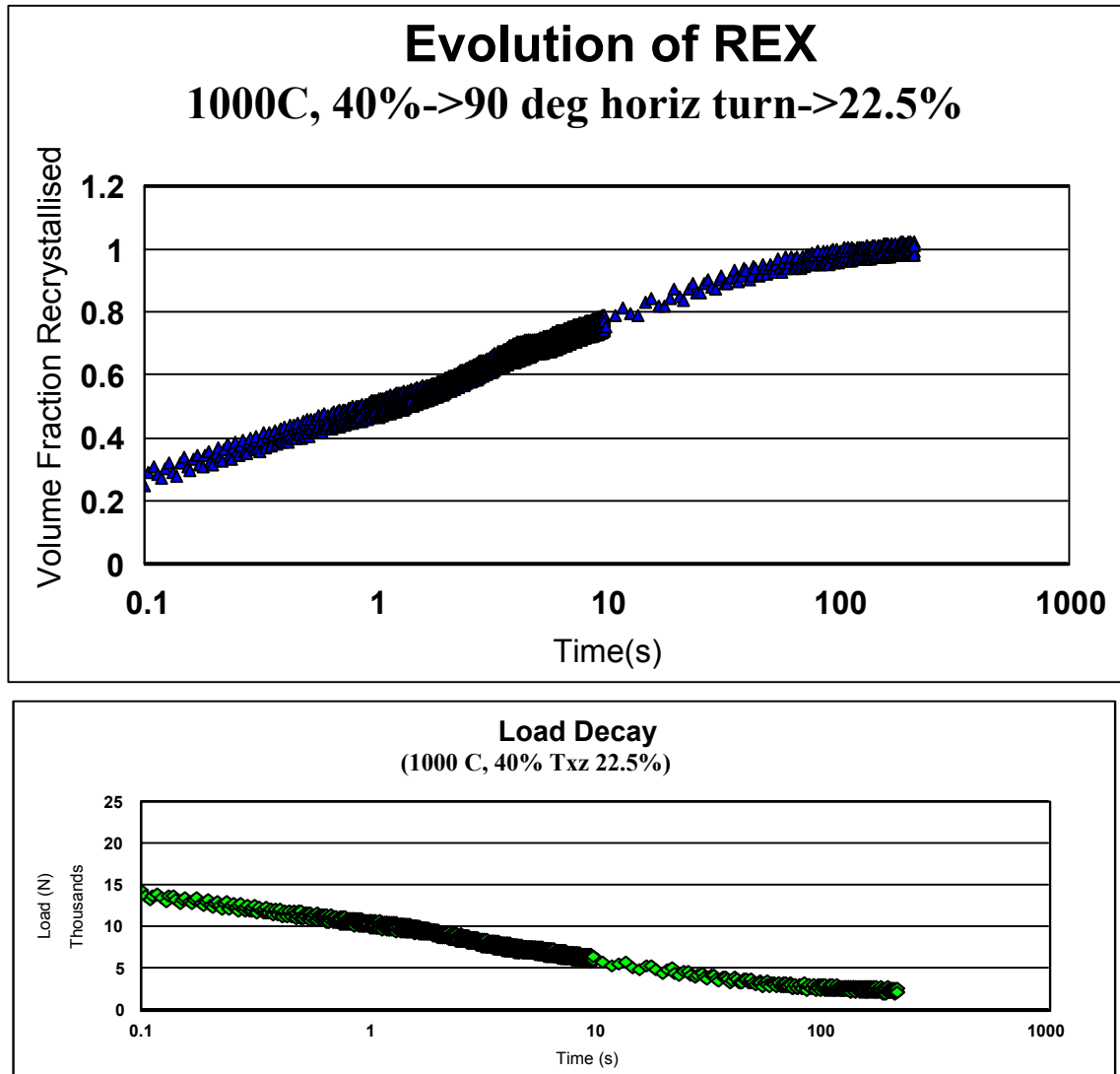


Fig. A1.30: Evolution of load and recrystallisation at 1000°C for compression of 40% followed by 22.5% with interpass horizontal turn of 90 degrees

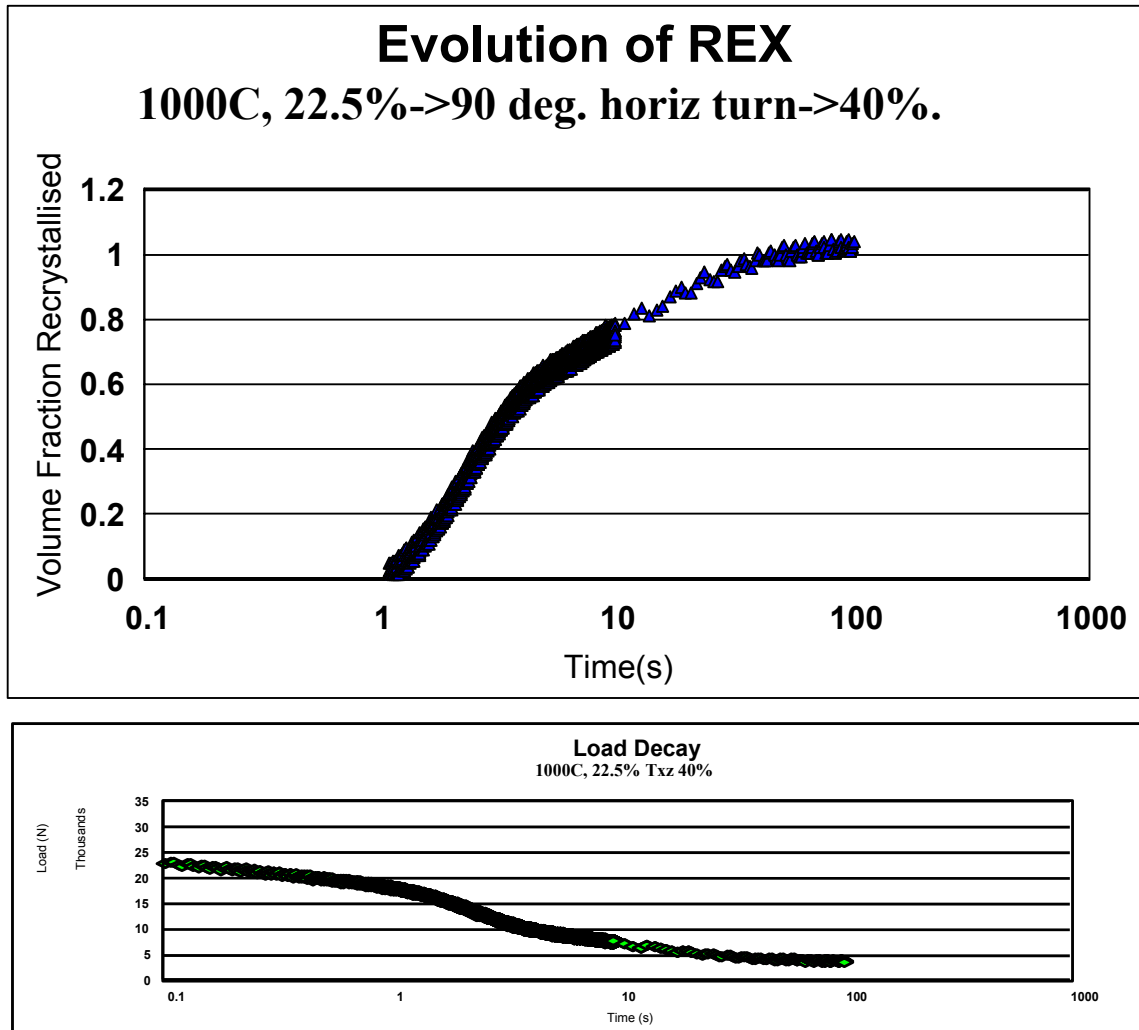


Fig. A1.31: Evolution of load and recrystallisation at 1000°C for compression of 22.5% followed by 40% with interpass horizontal turn of 90 degrees

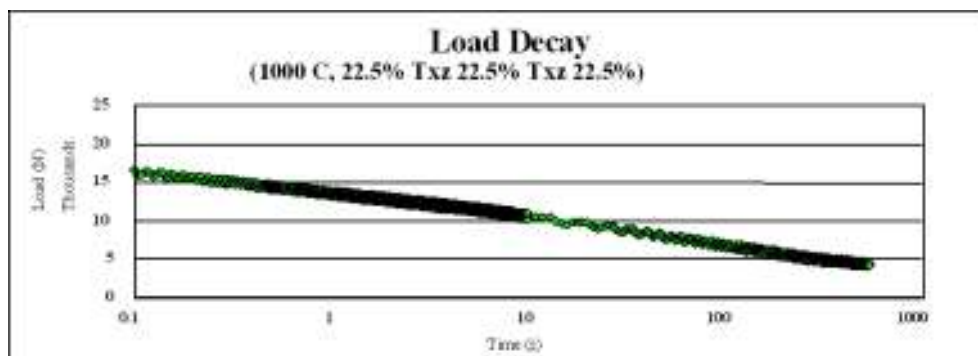


Fig. A1.32: Decay of load for triple hit of 22.5% at 1000° with 90 degree horizontal turn at each interpass

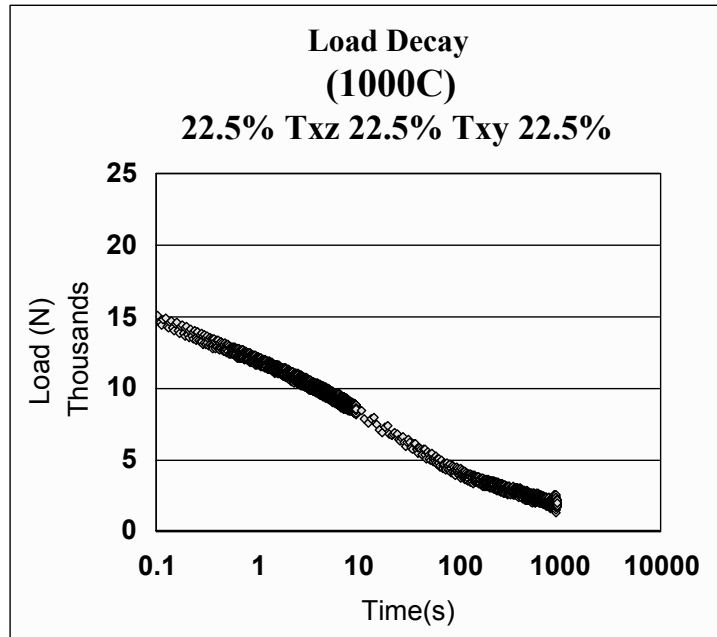
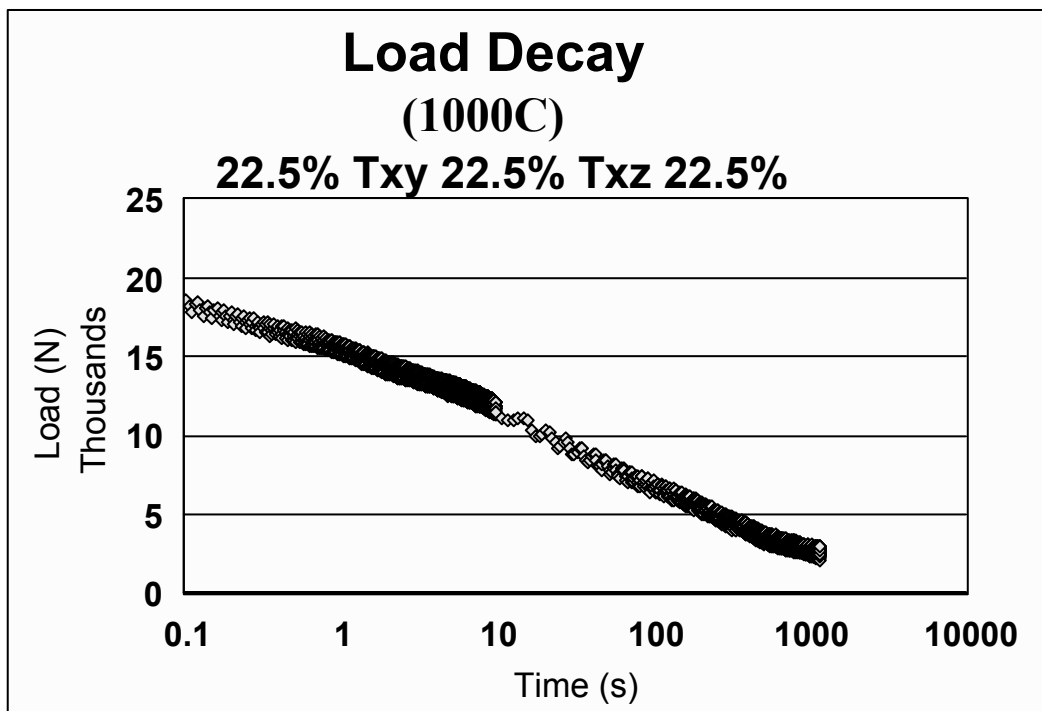
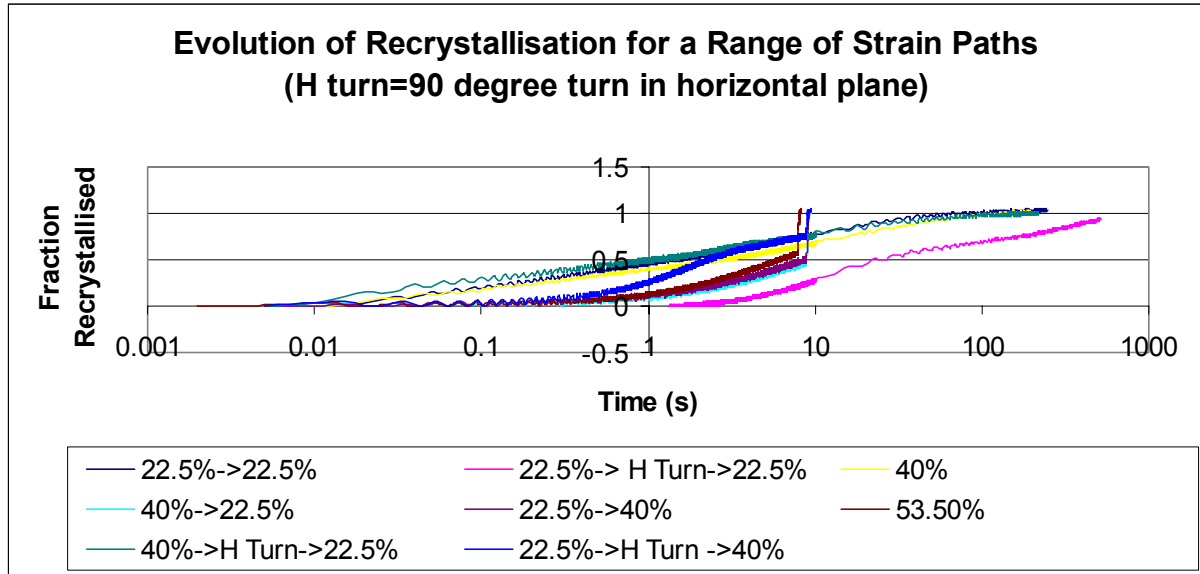


Fig. A1.33: Decay of load at 1100°C for triple hot of 22.5% with 90 degree turns during 1st and 2nd interpasses in horizontal and transverse/vertical planes respectively

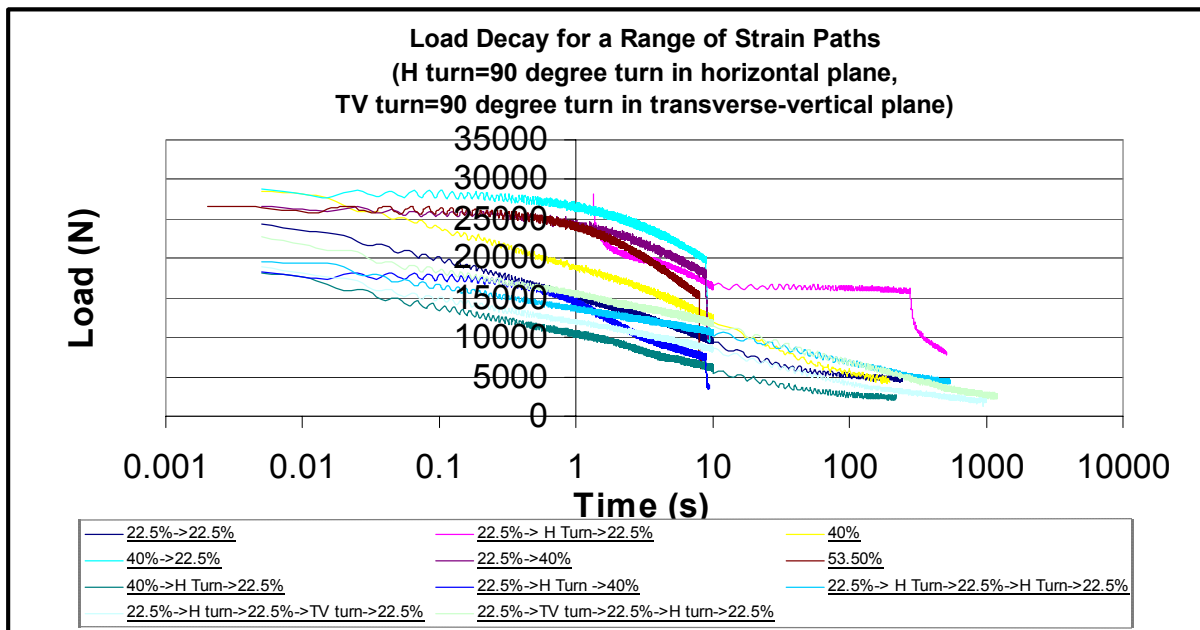


Recovery Only

Fig. A1.34: Decay of load at 1100°C for triple hit of 22.5% with 90 degree turns during 1st and 2nd interpasses in transverse/vertical and horizontal planes respectively



(a) Evolution of recrystallisation



(b) Load decay

Fig. A1.35(a and b): Evolution of recrystallisation and decay of load for range of strain paths at 1000°C

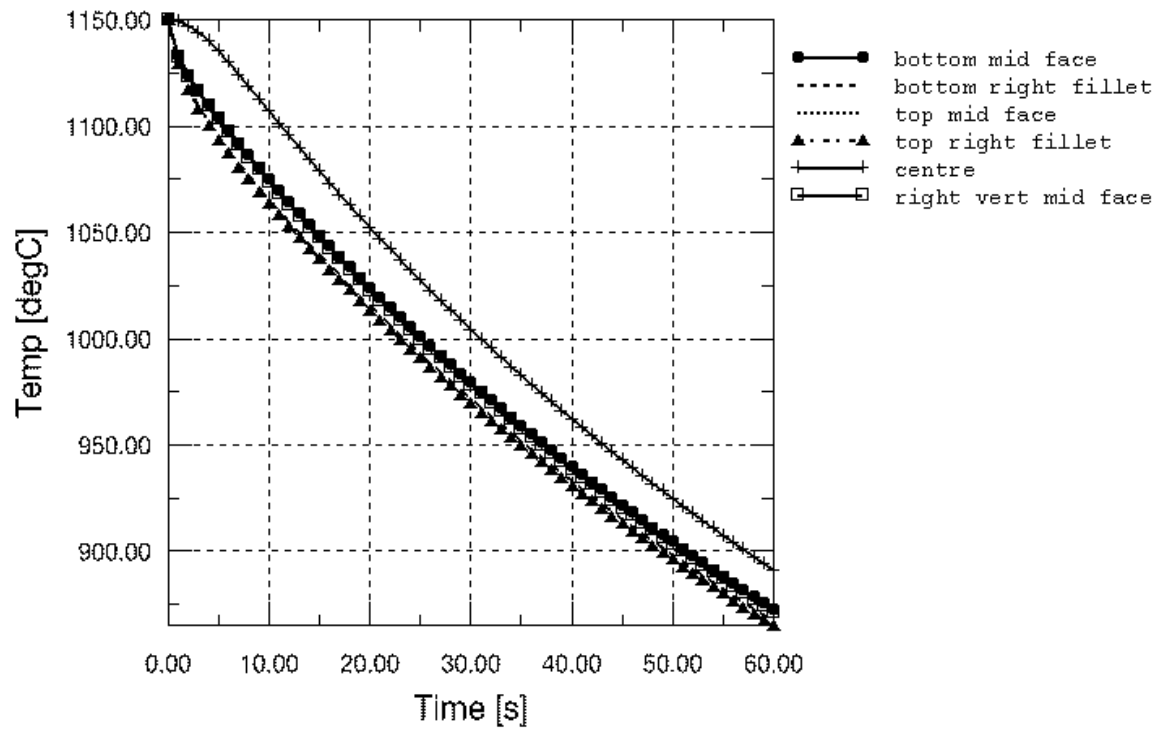


Fig. A1.36: Predicted cooling curves for selected points in Type 316 Stainless Steel specimen at Initial Uniform Temperature of 1150°C

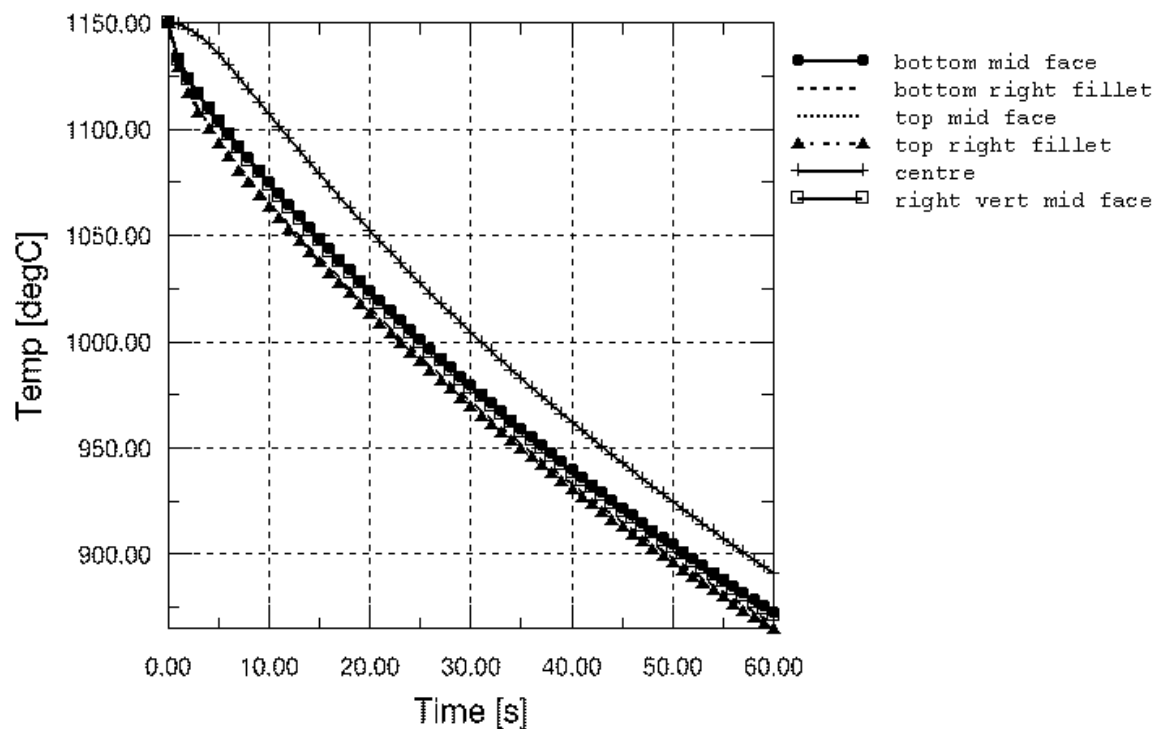


Fig. A1.37: Predicted cooling curves for selected points in Type 316 Stainless Steel specimen at Initial Uniform Temperature of 1120°C

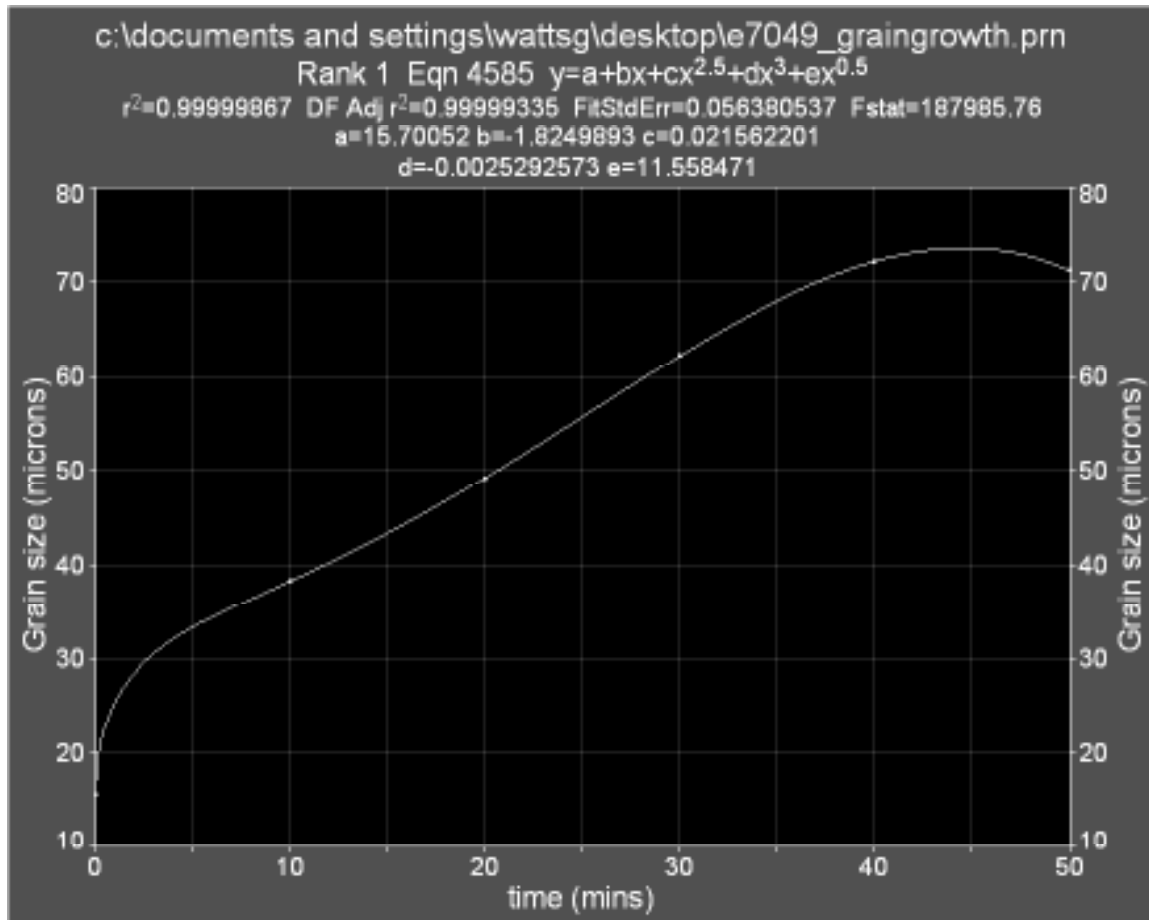
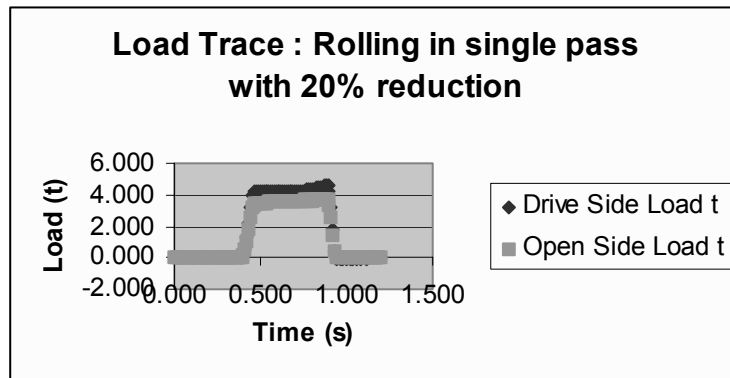
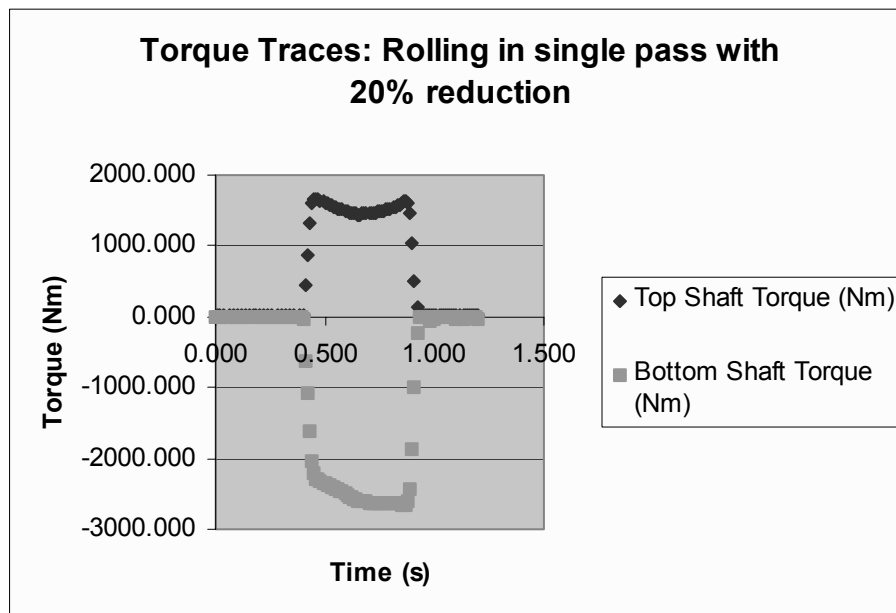


Fig. A1.38: Grain growth curve for type 316 stainless steel specimen held at 1100°C, fitted using TableCurve2D from measurements of grain size in samples held at temperature for times in the range 0 – 15 min

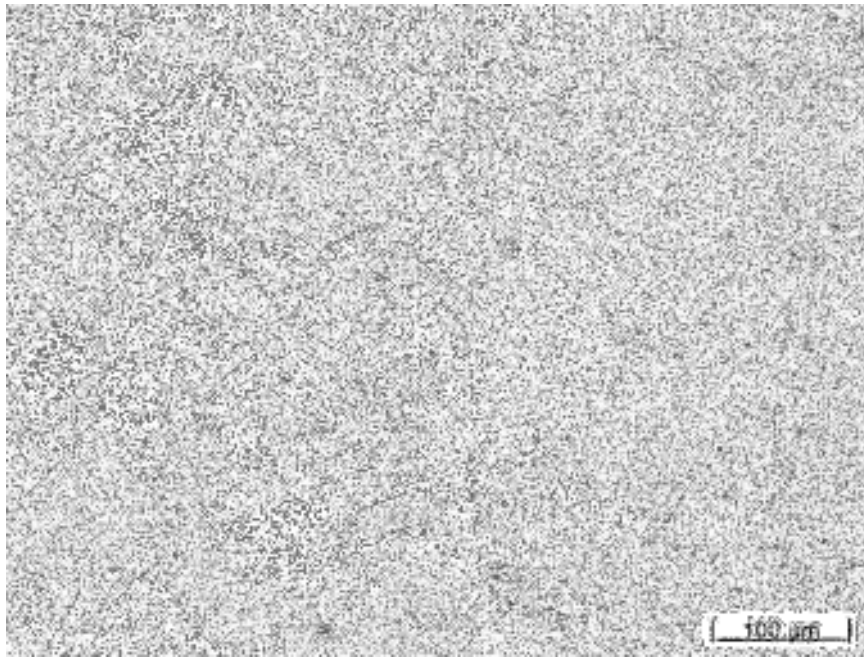


(a) Load traces

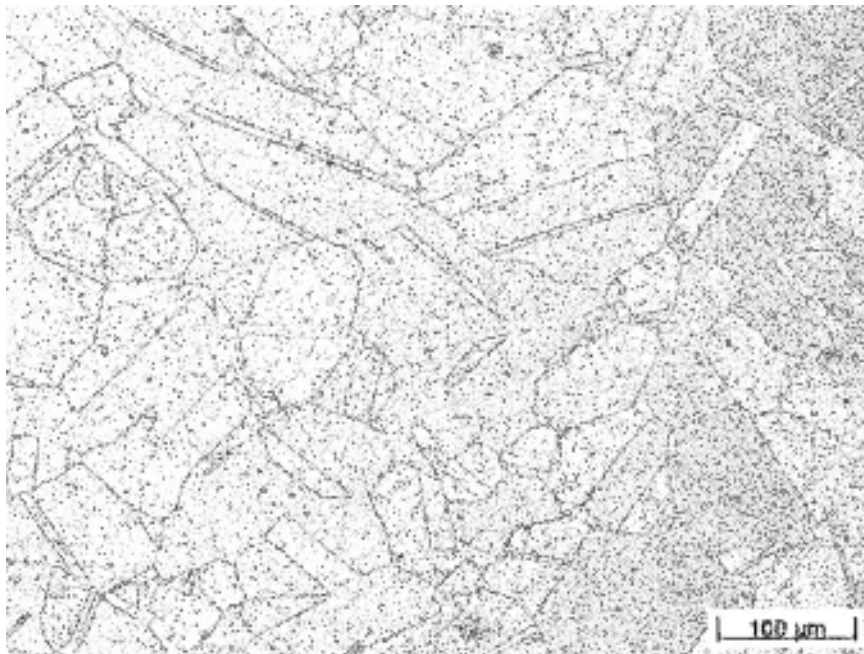


(b) Torque traces for top and bottom rolls

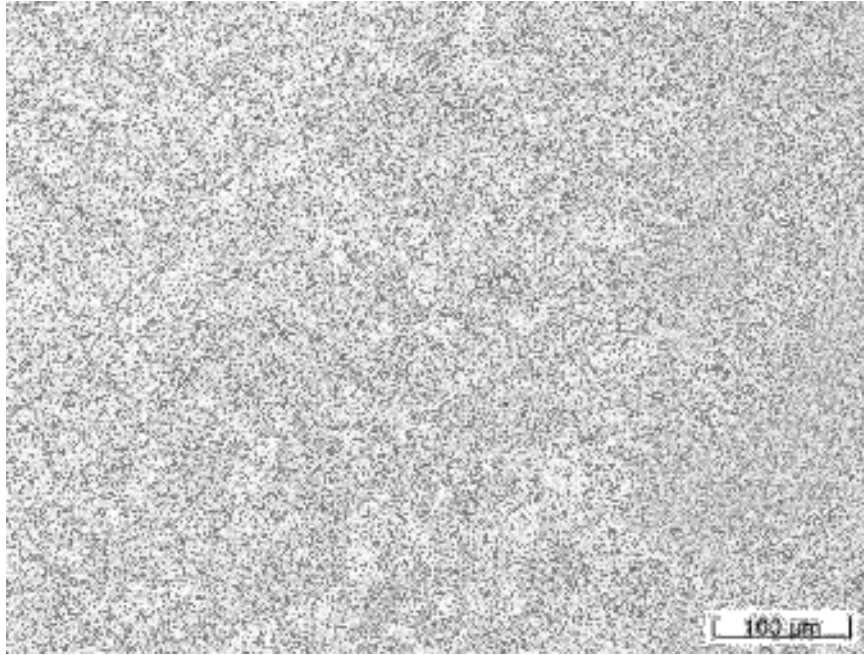
Fig. A1.39(a and b): Load and torque traces for rolling of type 316 stainless specimen in single pass (23 mm to 18.5 mm)



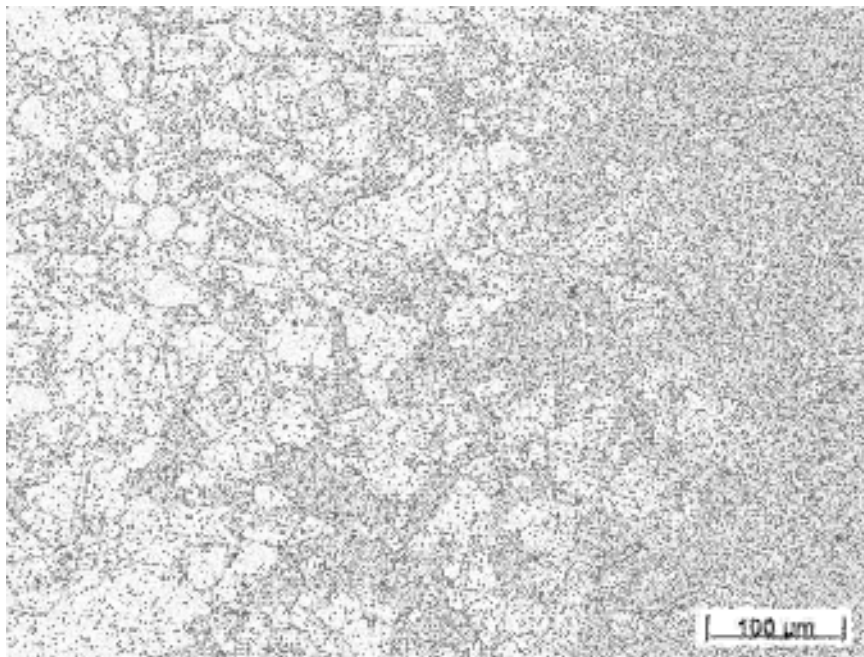
(a) $\frac{1}{4}$ depth position 1



(b) $\frac{1}{4}$ depth position 2

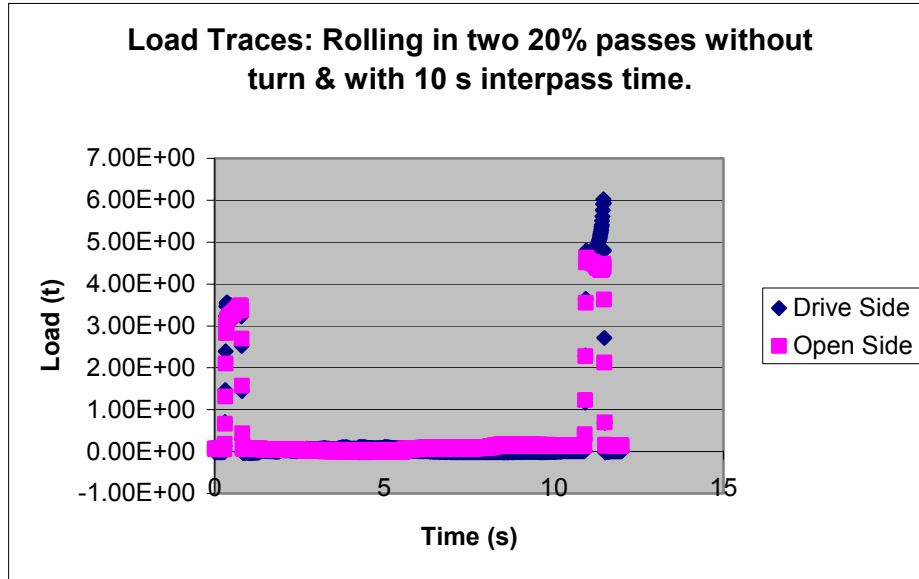


(c) Centre position 1

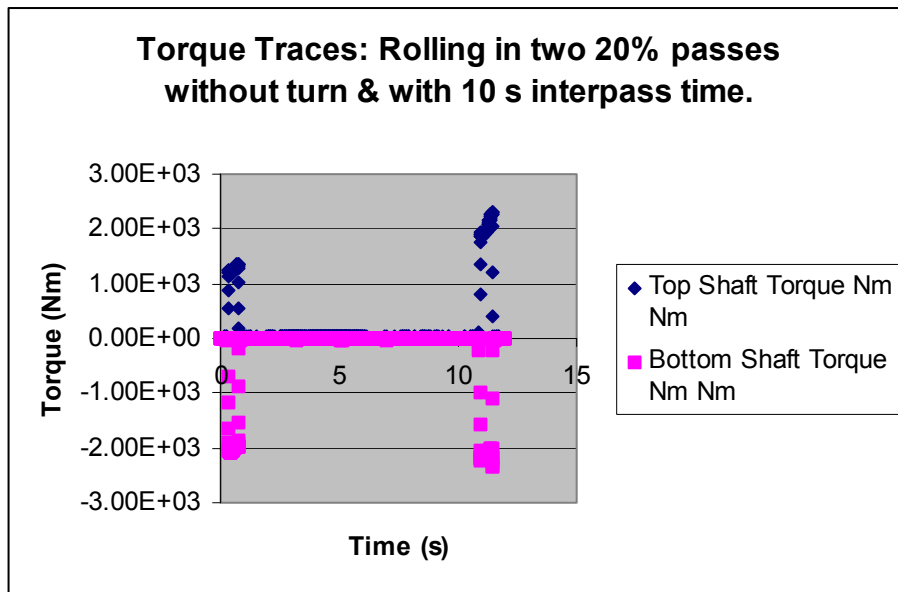


(d) Centre position 2

Fig. A1.40(a-d): Photographs of microstructure at $\frac{1}{4}$ depth and centre positions in sample 1 taken from type 316 stainless steel specimen 4R1F285 immediately after rolling in single pass from 23 mm to 18.5 mm thickness at 1100°C

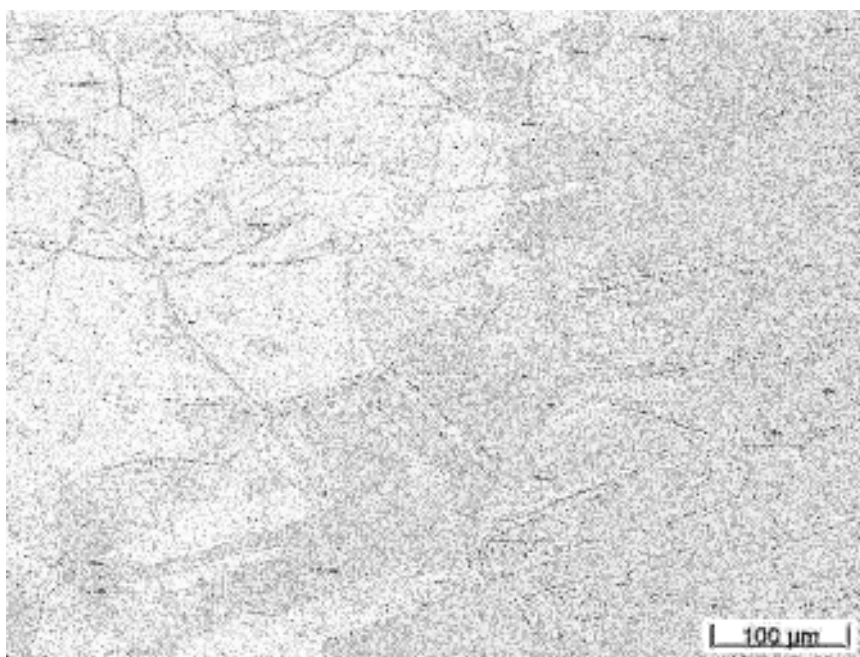


(a) Load

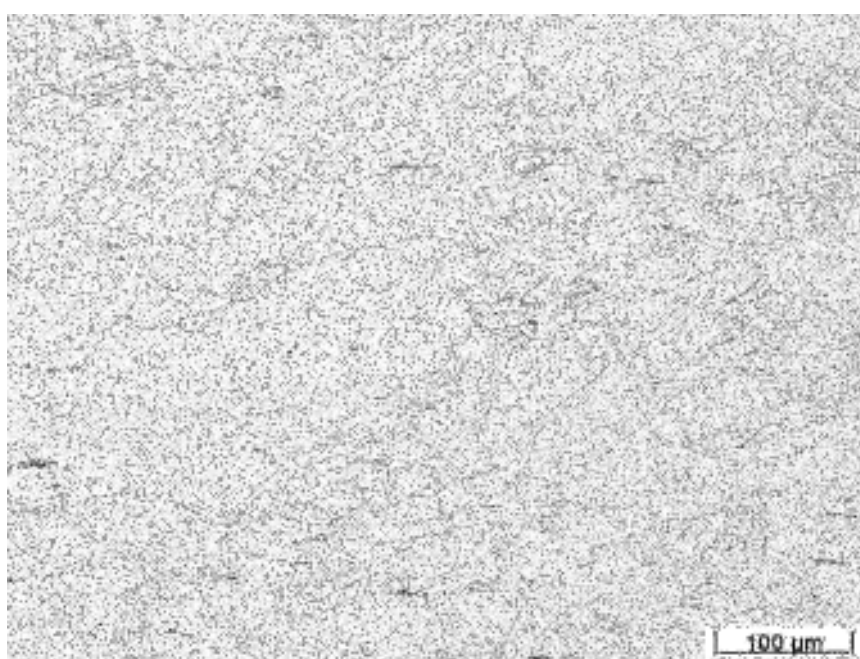


(b) Torque for top and bottom rolls

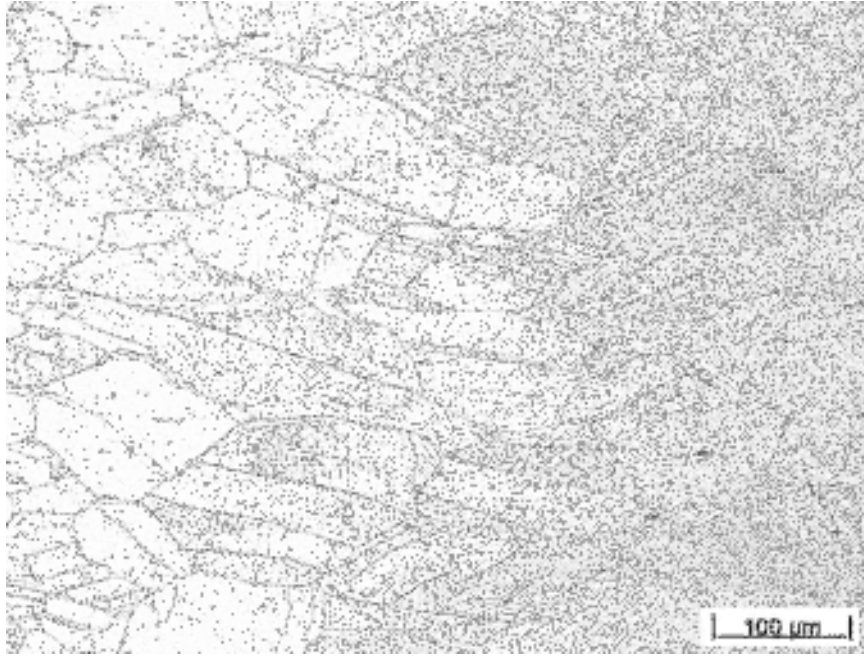
Fig. A1.41(a and b): Load and torque traces for rolling of stainless specimen in 2 passes (23 mm to 18.5 mm to 15 mm) with a 10 s interpass time and no turning



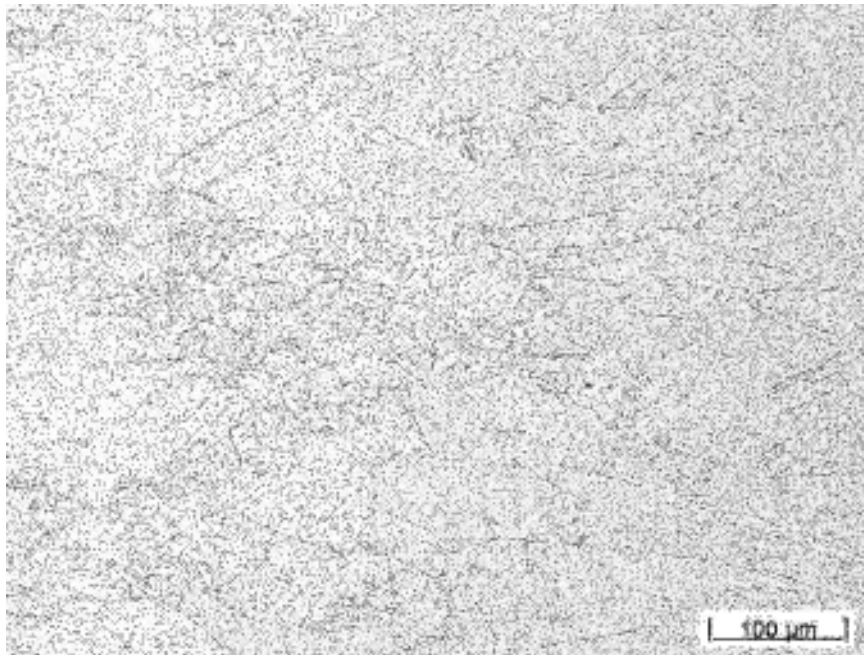
(a) Unreheated, at $\frac{1}{4}$ depth



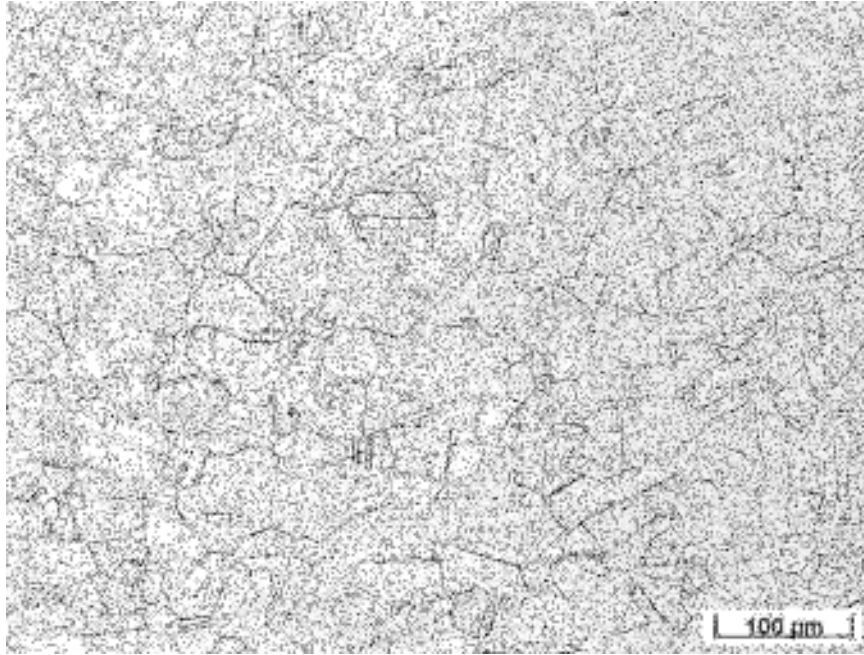
(b) Unreheated, at centre position



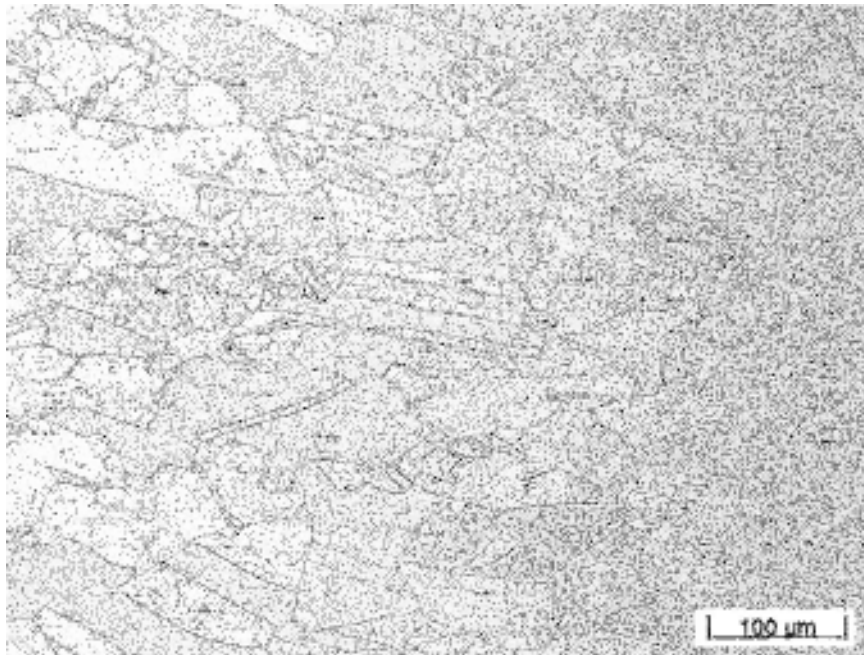
(c) Reheated for 5 s, at $\frac{1}{4}$ depth



(d) Reheated for 5 s, at centre position

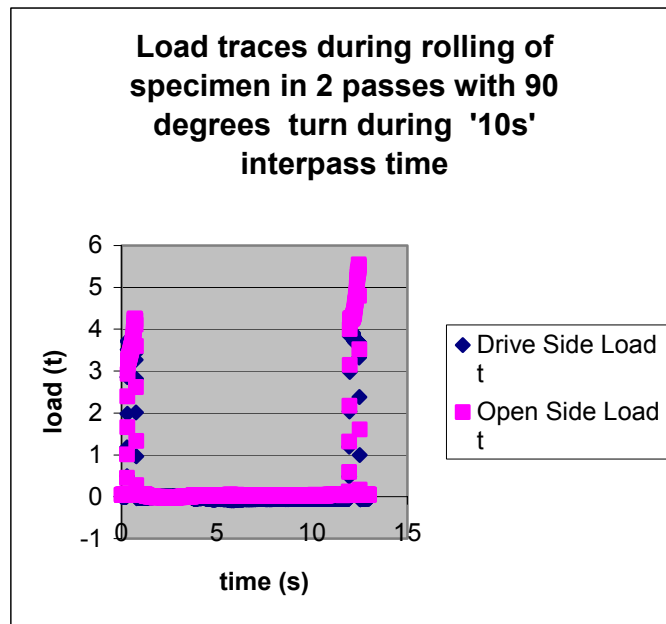


(e) Reheated for 60 s at $\frac{1}{4}$ depth

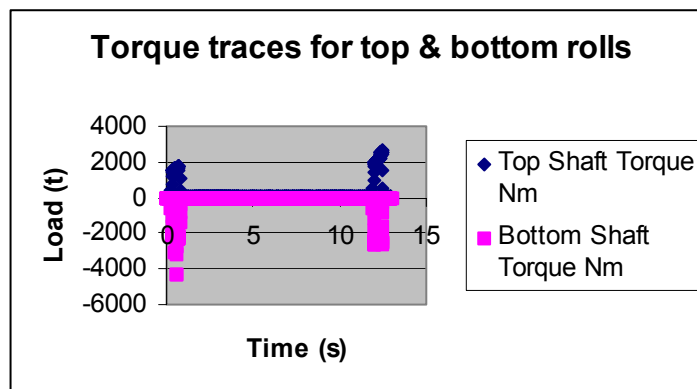


(f) Reheated for 60 s, at centre position

Fig. A1.42(a-f): Photographs of microstructure at $\frac{1}{4}$ depth and centre positions in samples from transverse sections A, B and C taken from type 316 stainless steel specimen 4R1F287 immediately after rolling in two 20% passes (interpass '10' s, 1100°C, no turn) and reheated respectively for 0, 5 and 60 s (test reports K4M3011, K4M0312)

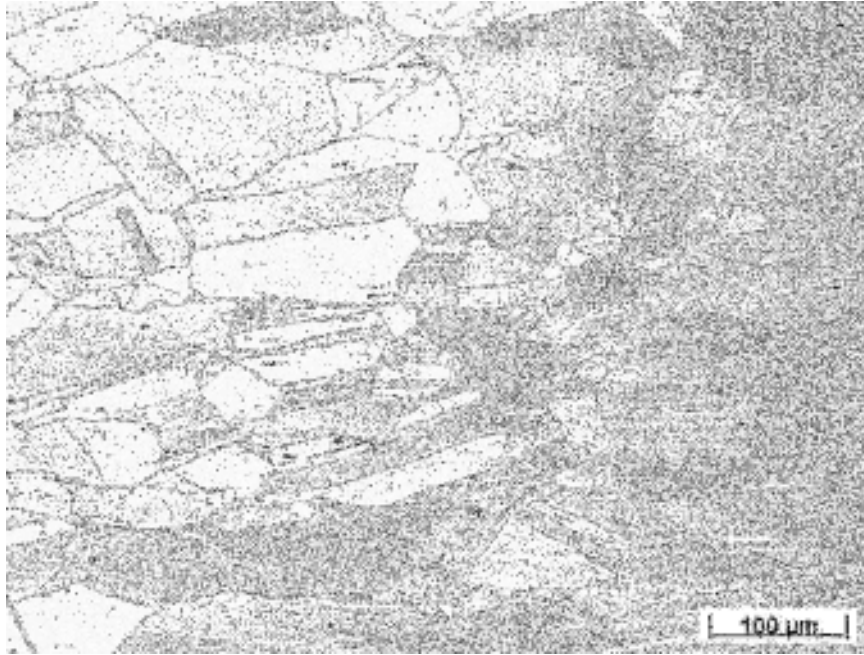


(a) Load

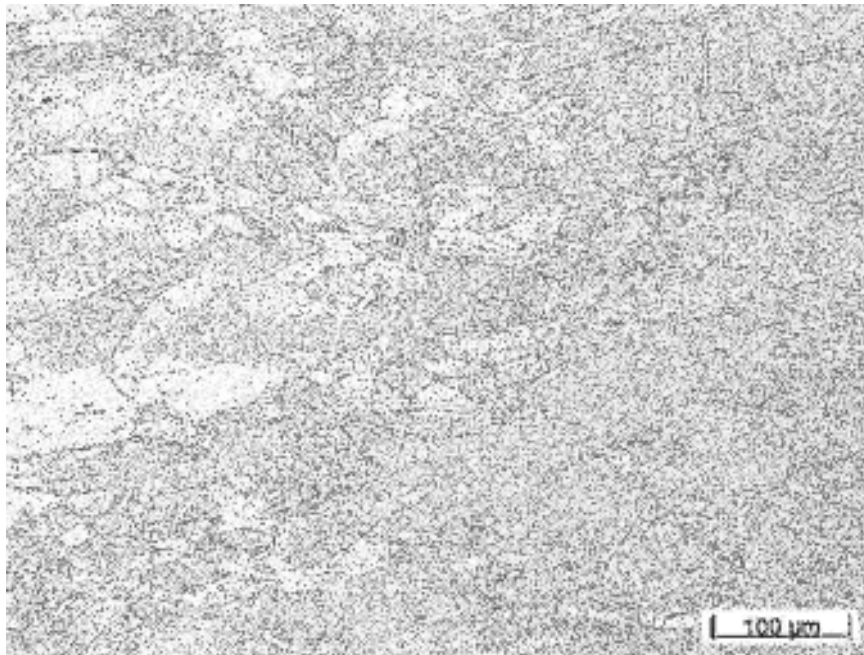


(b) Torque traces

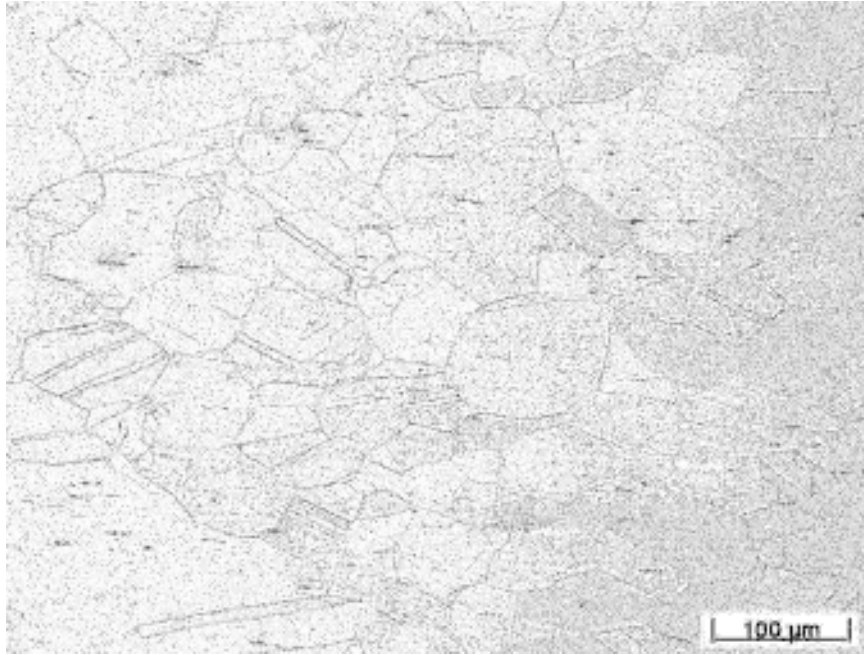
Fig. A1.43(a and b): Load and torque traces for rolling of type 316 stainless steel specimen in two 20% passes with 90° turning in '10 s' interpass time



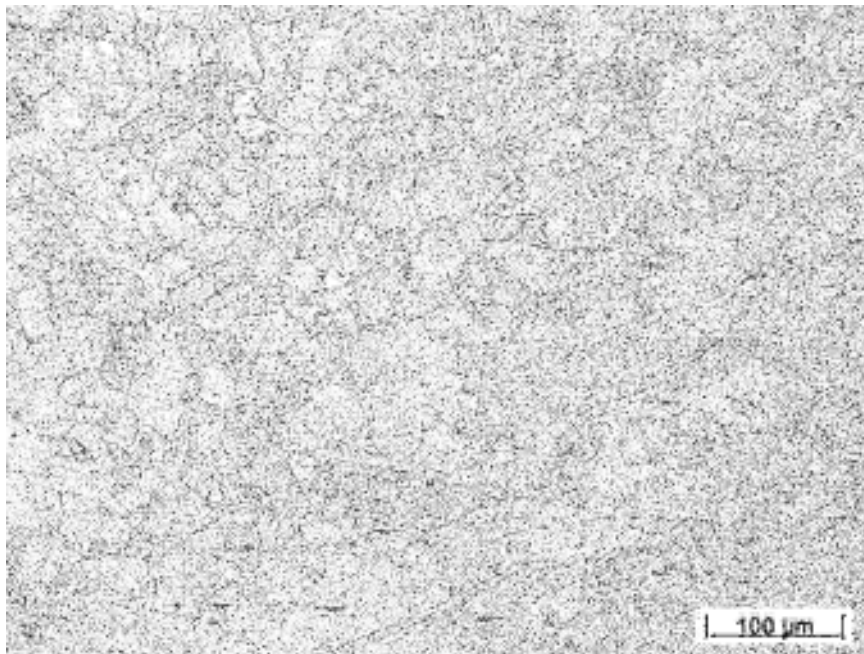
(a) Unreheated, sample from location 1, at $\frac{1}{4}$ depth



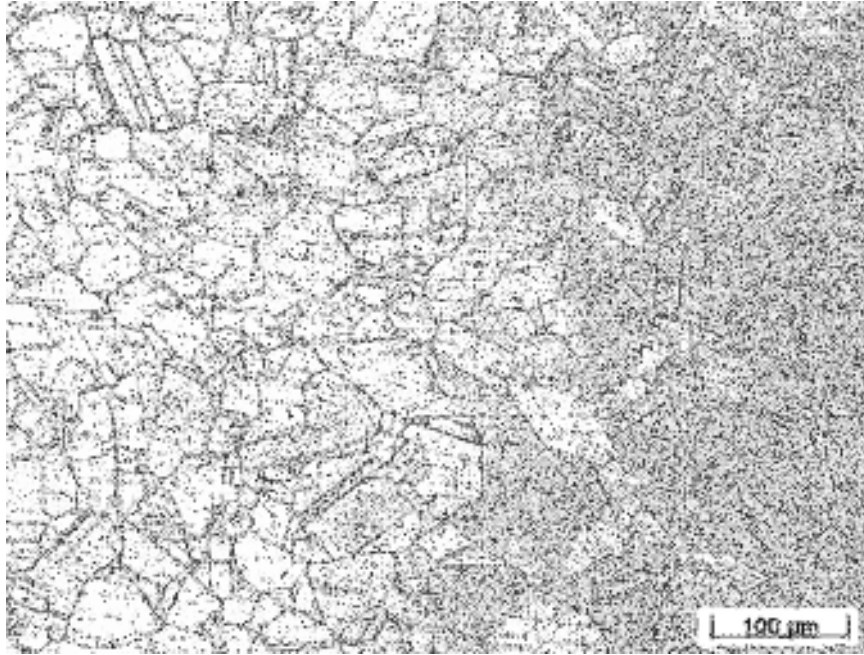
(b) Unreheated, sample from location 1, at centre position



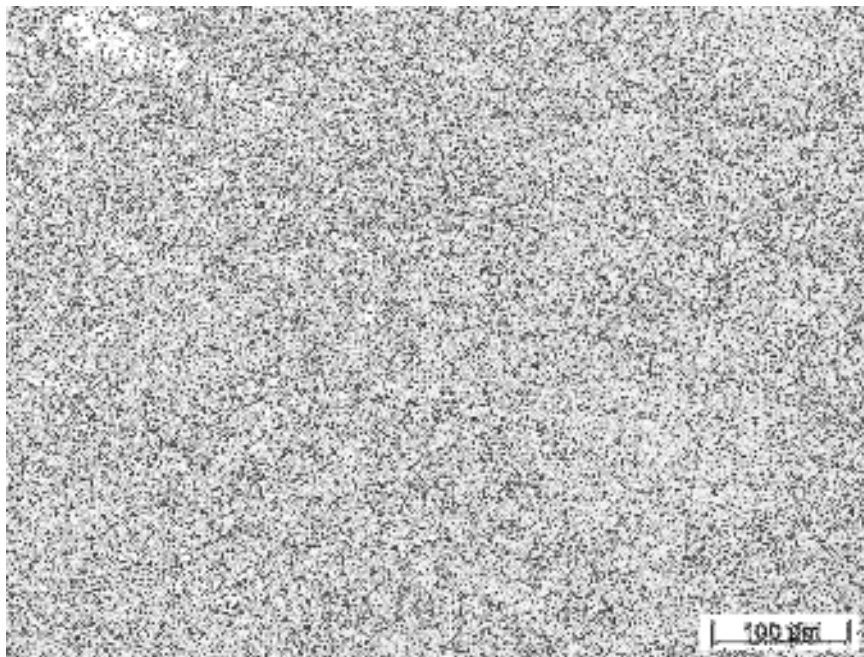
(c) Unreheated, sample from location 2, at $\frac{1}{4}$ depth



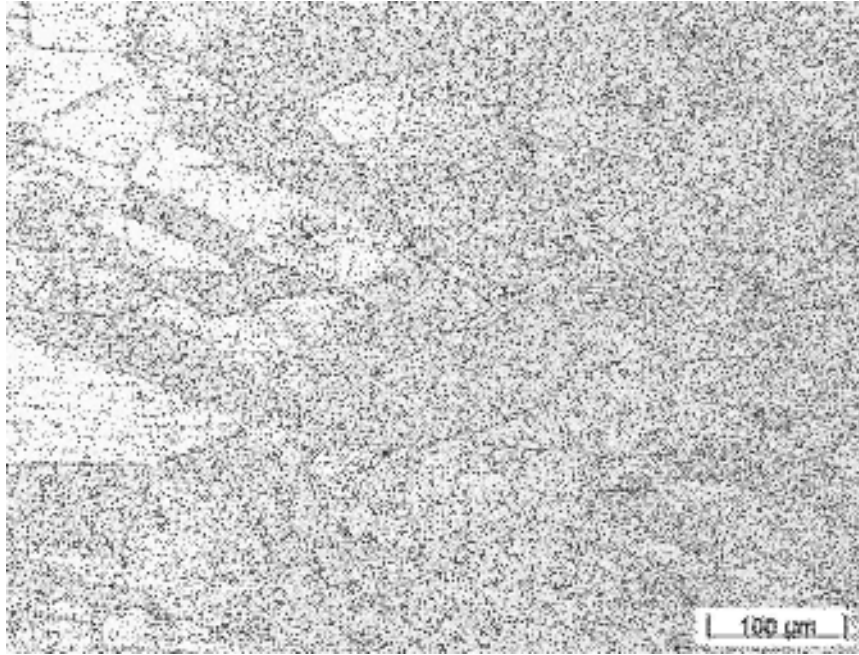
(d) Unreheated, sample from location 2, at centre position



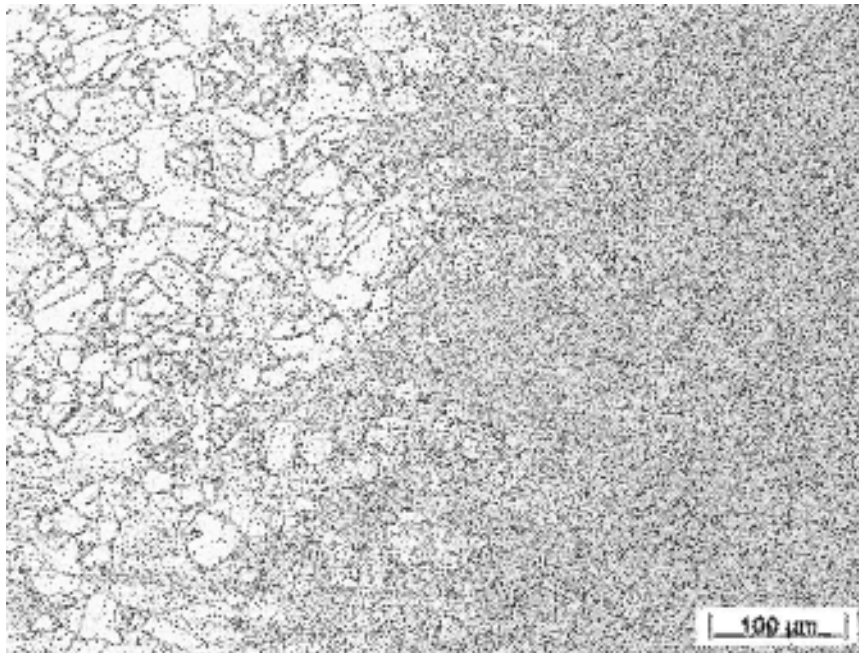
(e) Reheated for 5 s, sample from location 1, at $\frac{1}{4}$ depth



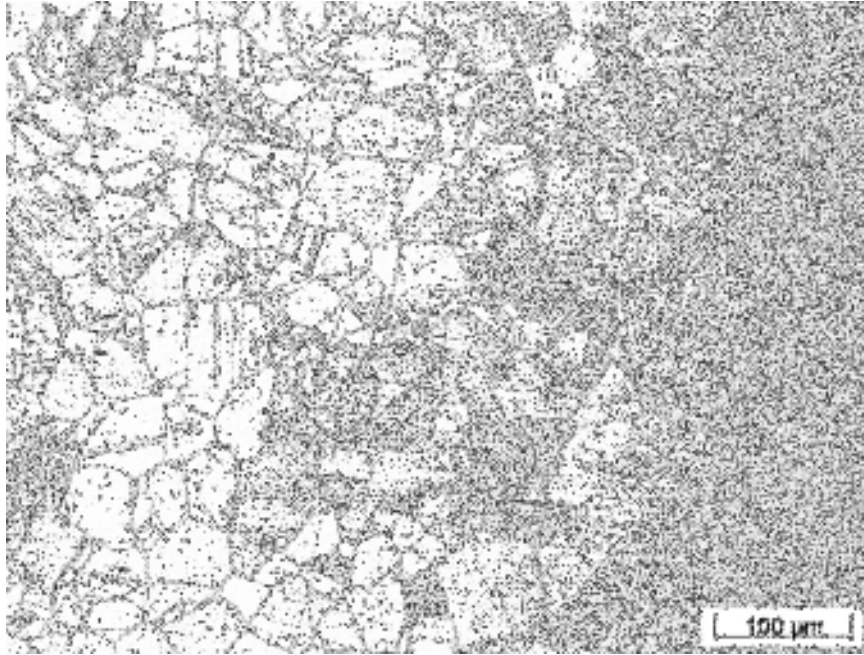
(f) Reheated for 5 s, sample from location 1, at centre position



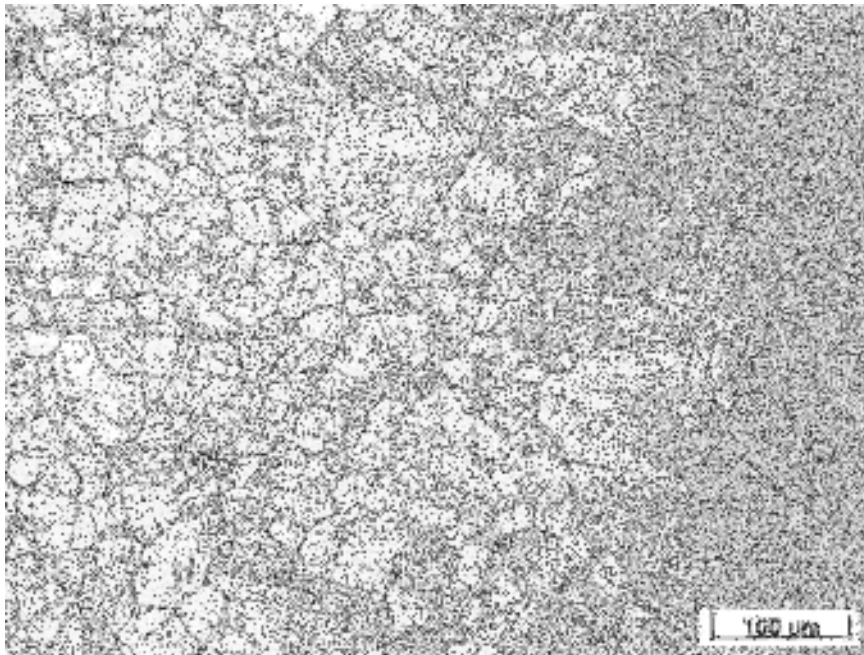
(g) Reheated for 5 s, sample from location 2, at $\frac{1}{4}$ depth



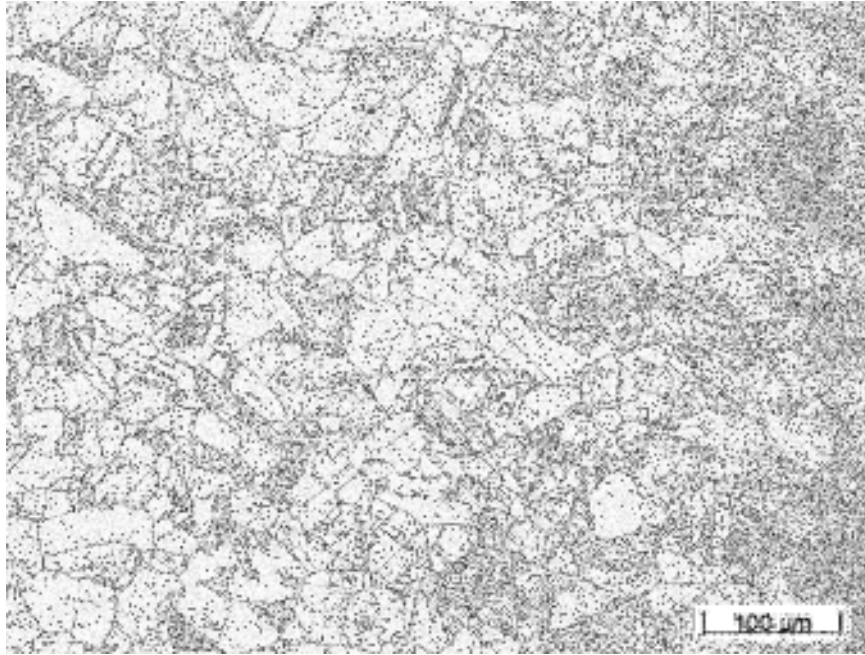
(h) Reheated for 5 s, sample from location 2, at centre position



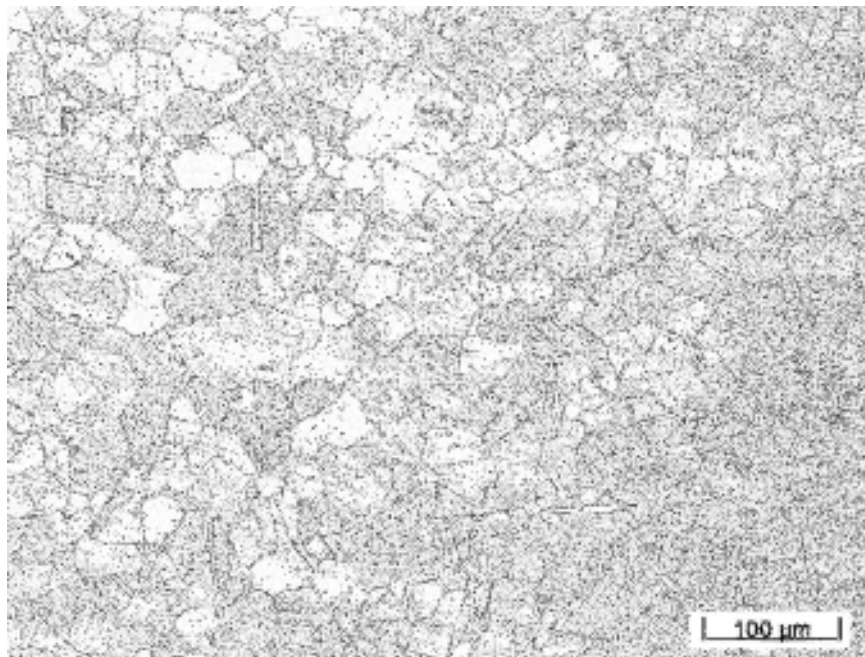
(i) Reheated for 60 s, sample from location 1, at $\frac{1}{4}$ depth



(j) Reheated for 60 s, sample from location 1, at centre position

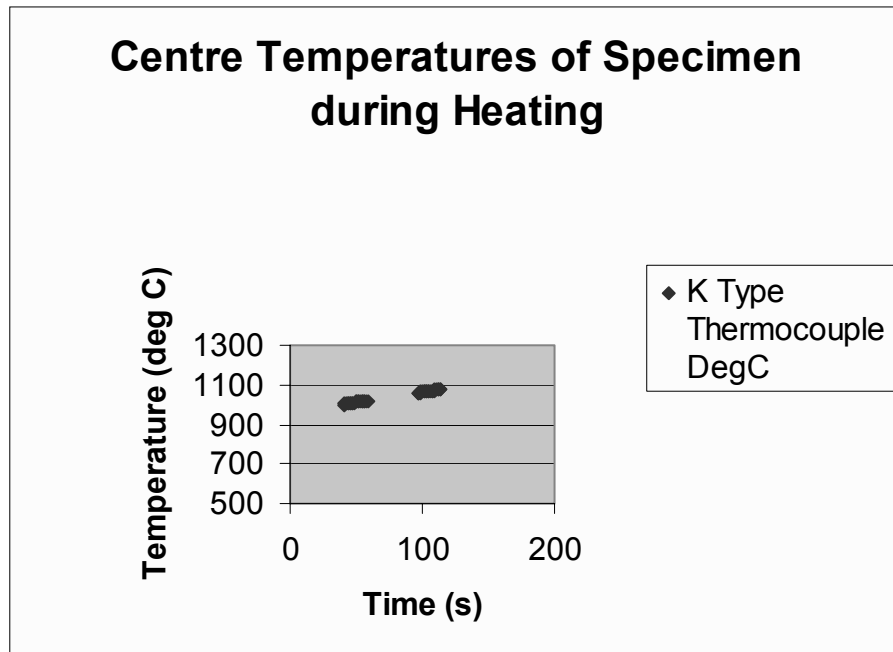


(k) Reheated for 60 s, sample from location 2, at $\frac{1}{4}$ depth

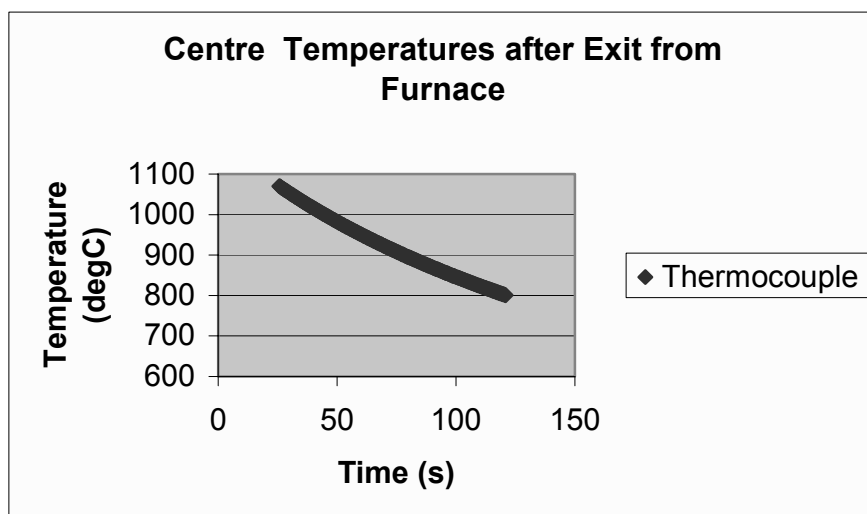


(l) Reheated for 60 s, sample from location 2, at centre position

Fig. A1.44(a-l): Photographs of microstructure at $\frac{1}{4}$ depth and centre positions in samples, at locations (1) perpendicular and (2) parallel to the rolled surface, in transverse sections A, B and C taken from type 316 stainless steel specimen 4R1F287 immediately after rolling in two 20% passes with 90° turn ('10 s' interpass, temperature 1100°C) and reheated respectively for 0, 5 and 60 s (test reports K4M0312



(a) Reheating



(b) Cooling

Fig. A1.45(a and b): Heating and cooling curves for points at centre and surface of specimen obtained from thermocouple readings during residence in and exit from reheating furnace

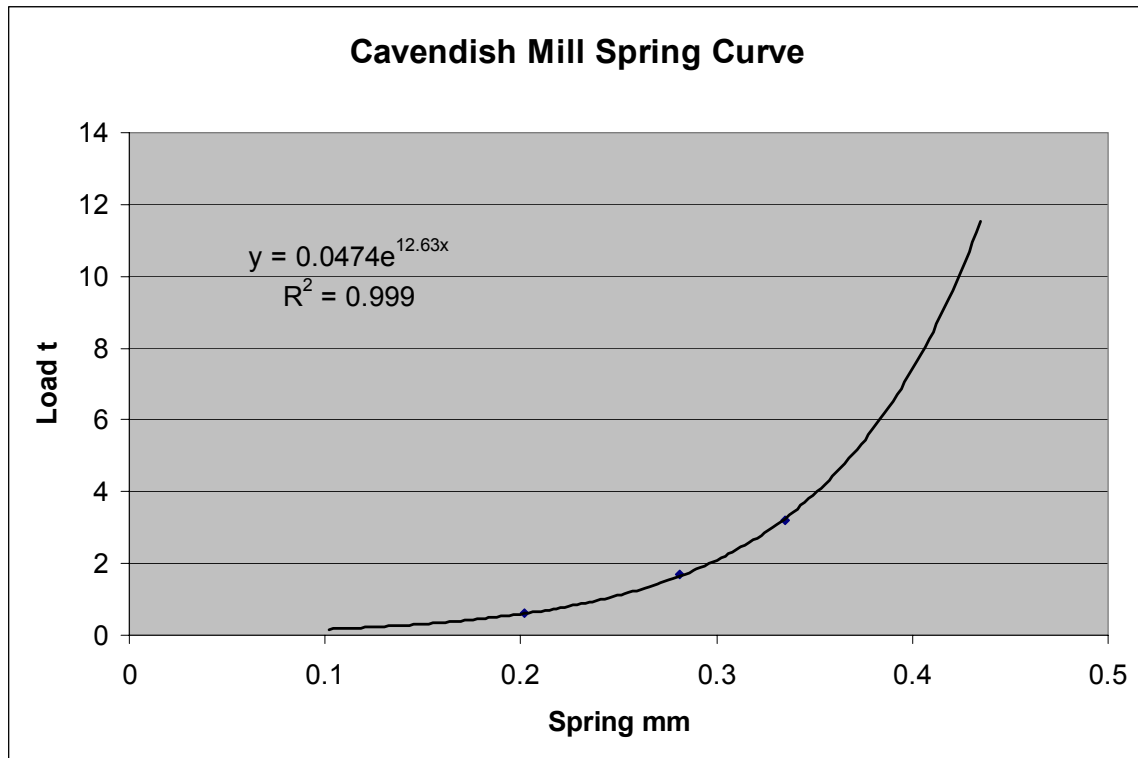
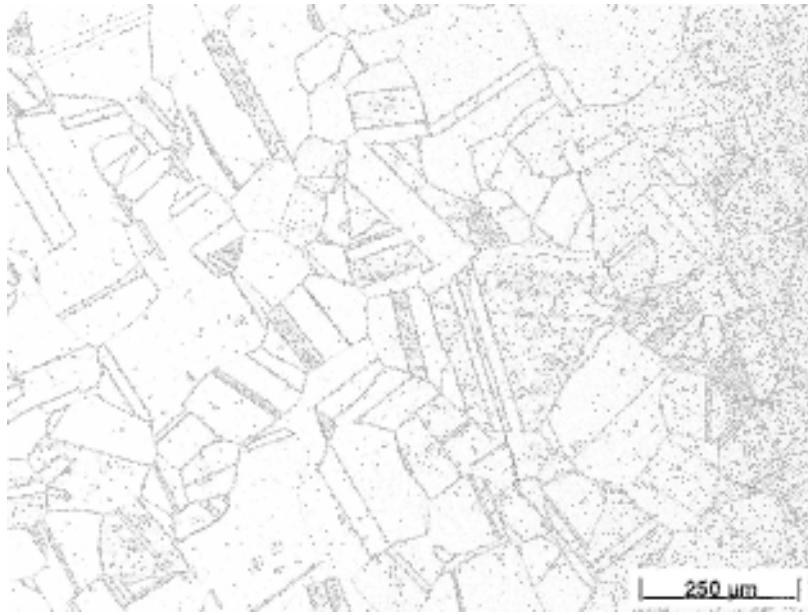
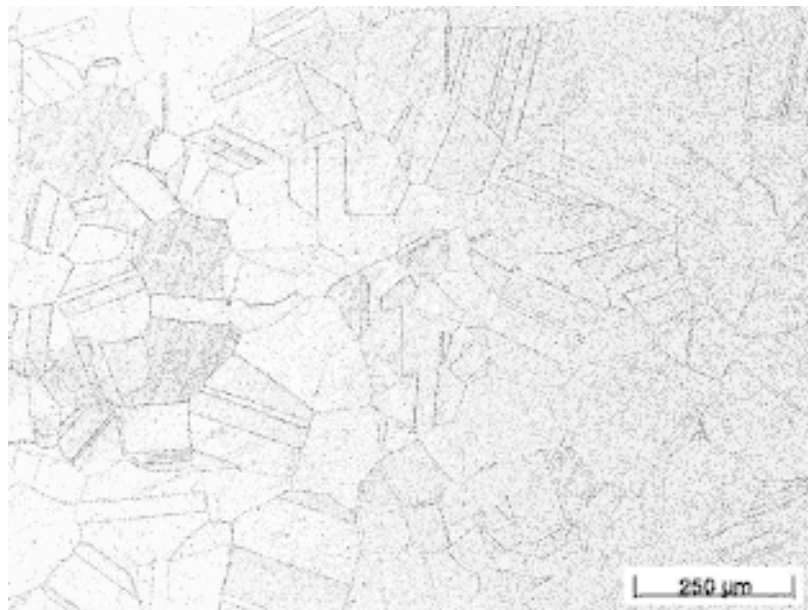


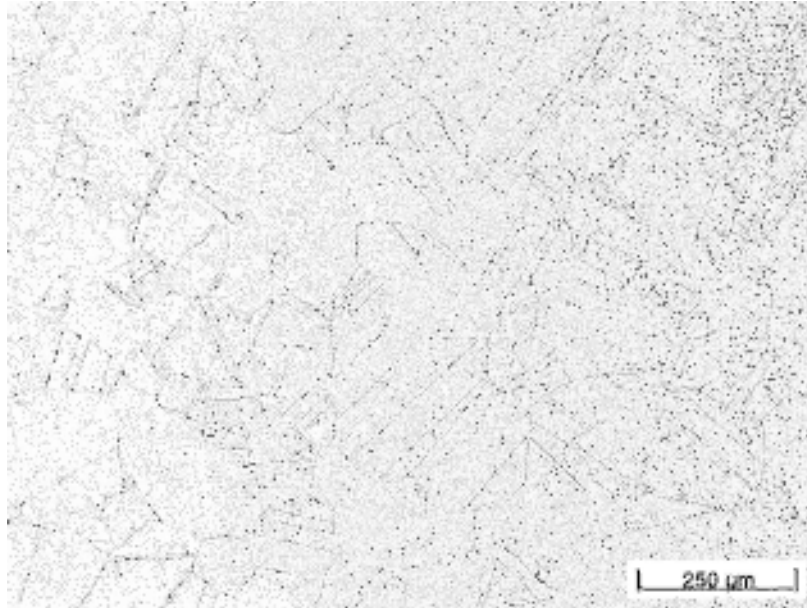
Fig A1.46: Spring curve for Cavendish Rolling Mill
(see Tech Report 6)



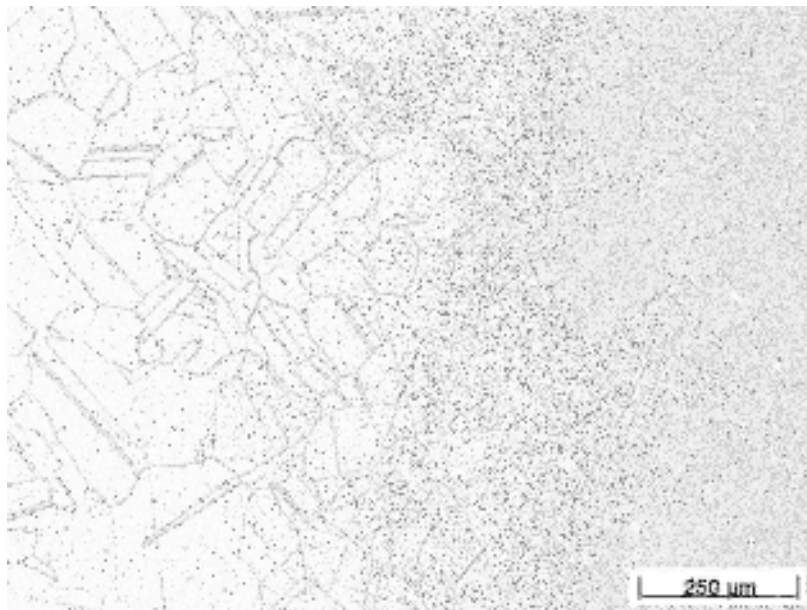
(a)



(b)



(c)



(d)

Fig. A1.47(a-d): Micrographs of homogenised Type 316 Stainless Steel samples 4R1F64B, 4R1F71B, 4R1F284B and 4R1f514B (as in Test Report No. K4M0340)

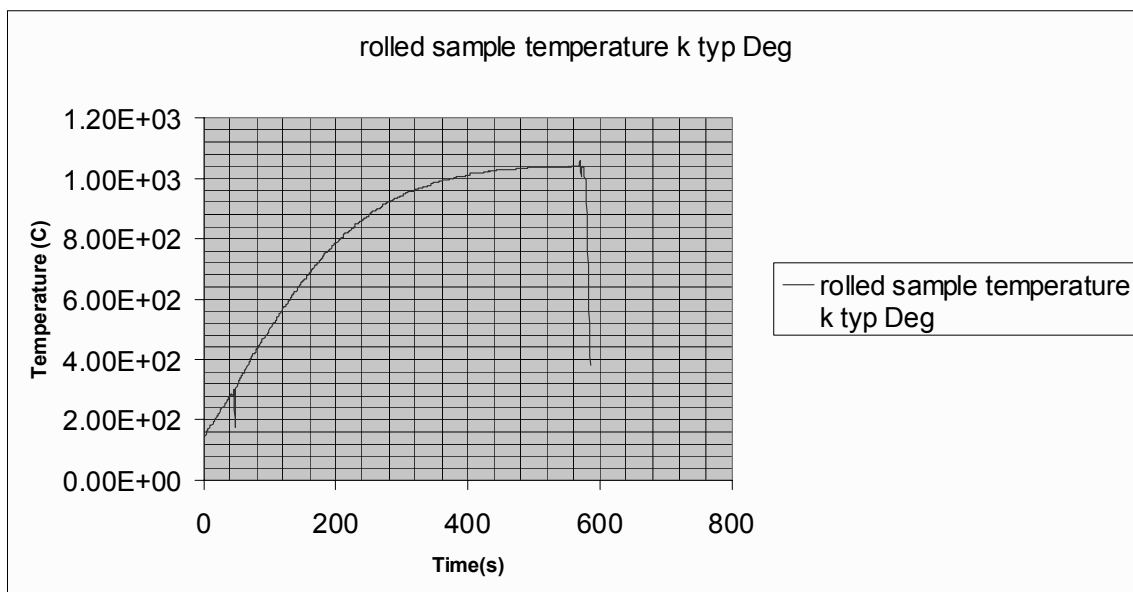


Fig. A1.48: Temperature profile employed prior to pass 2 when reheating to 1040°C and holding for 60 s

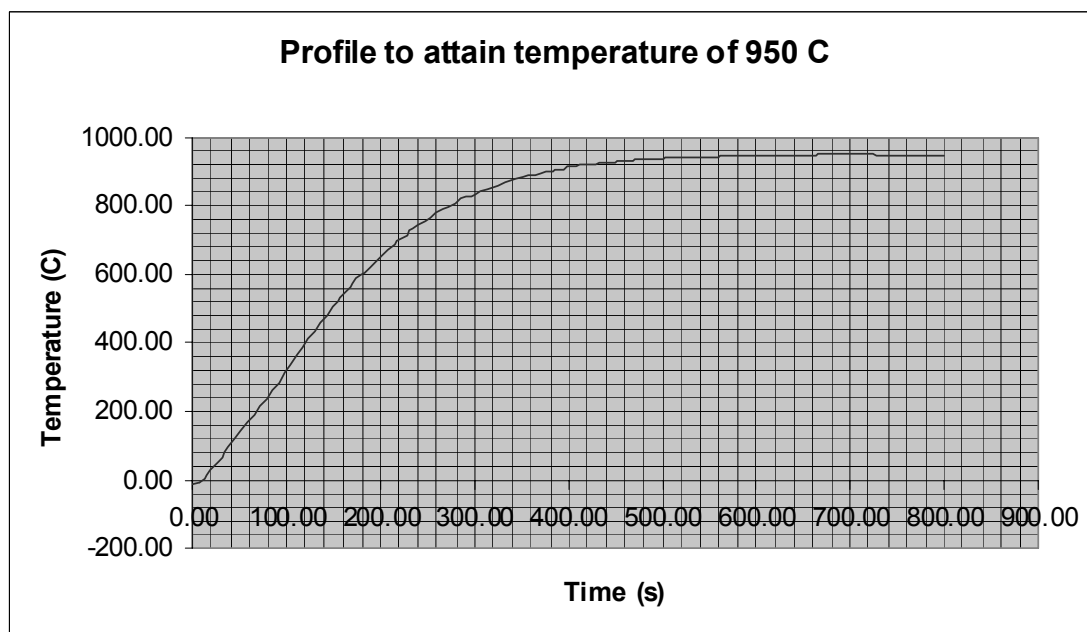
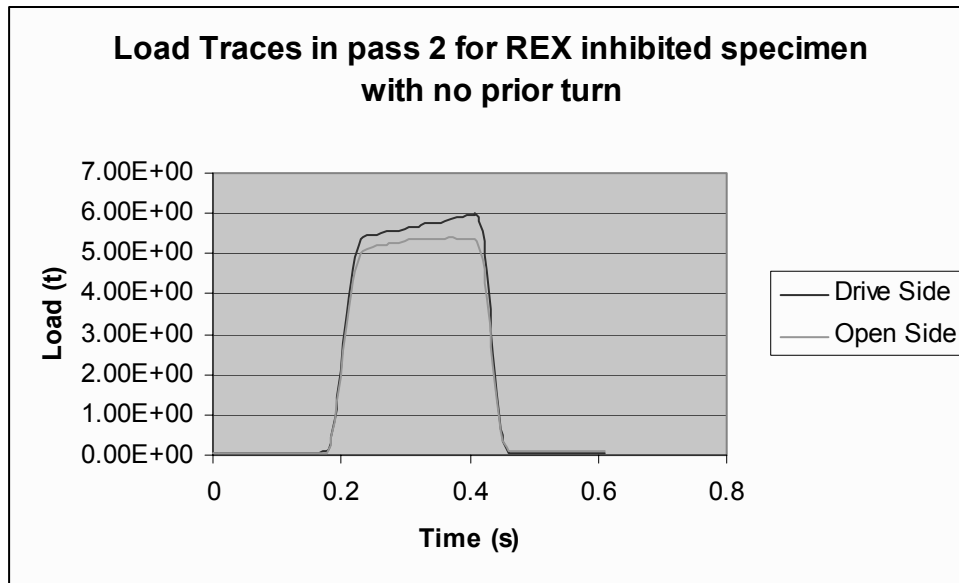
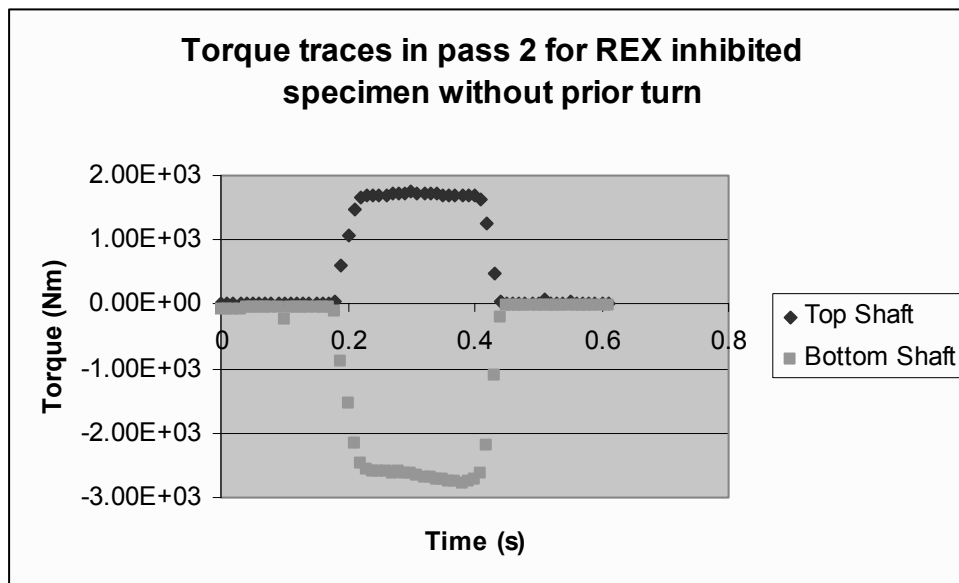


Fig. A1.49: Temperature Profile employed when reheating to 950°C prior to pass 2 (without turn)

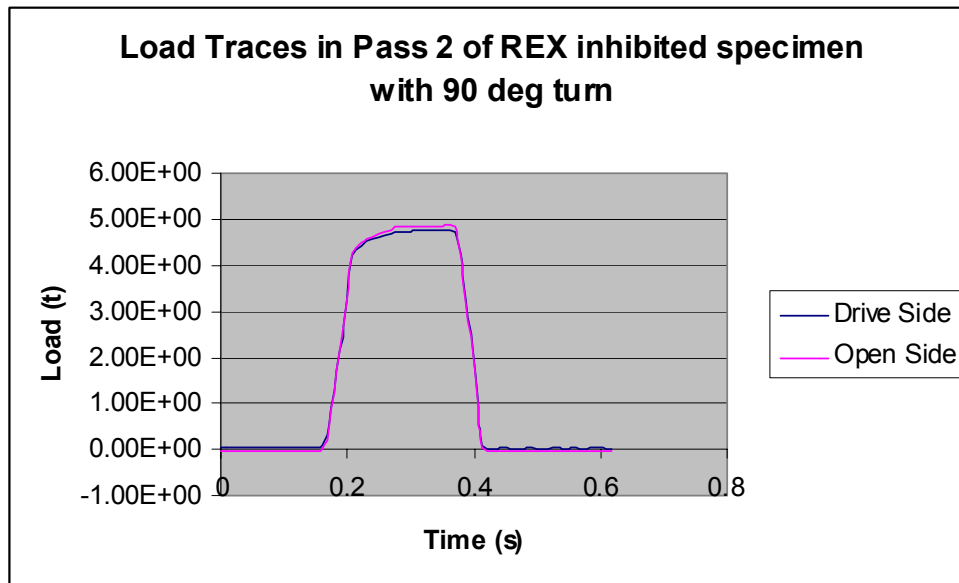


(a)

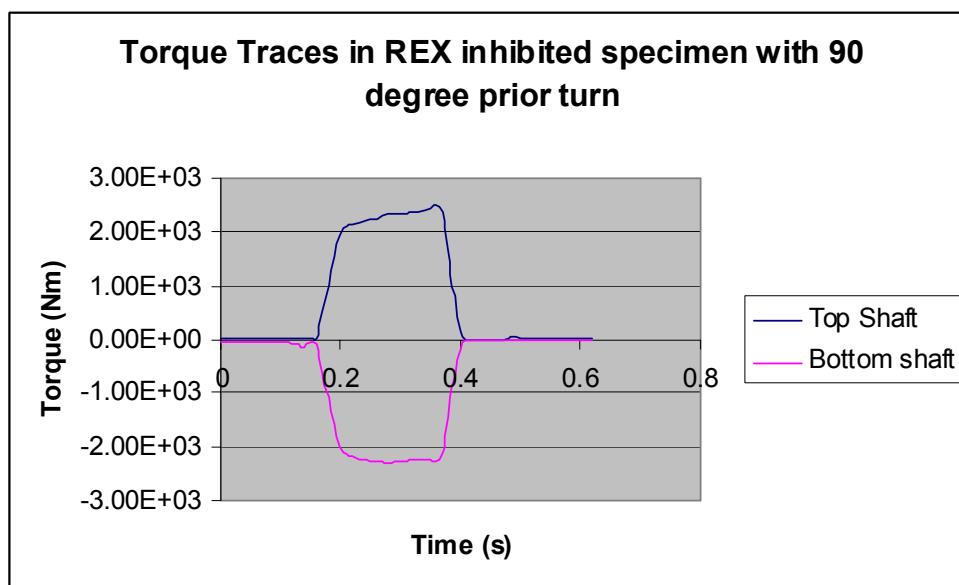


(b)

Fig. A1.50(a and b): Load and torque traces in Pass 2 for specimen between passes given single reheat to 950°C

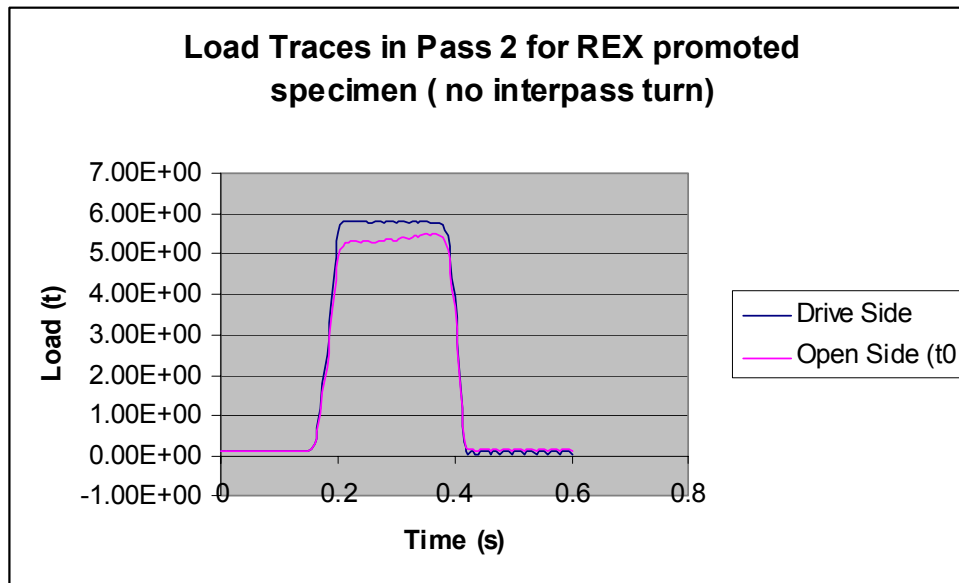


(a)

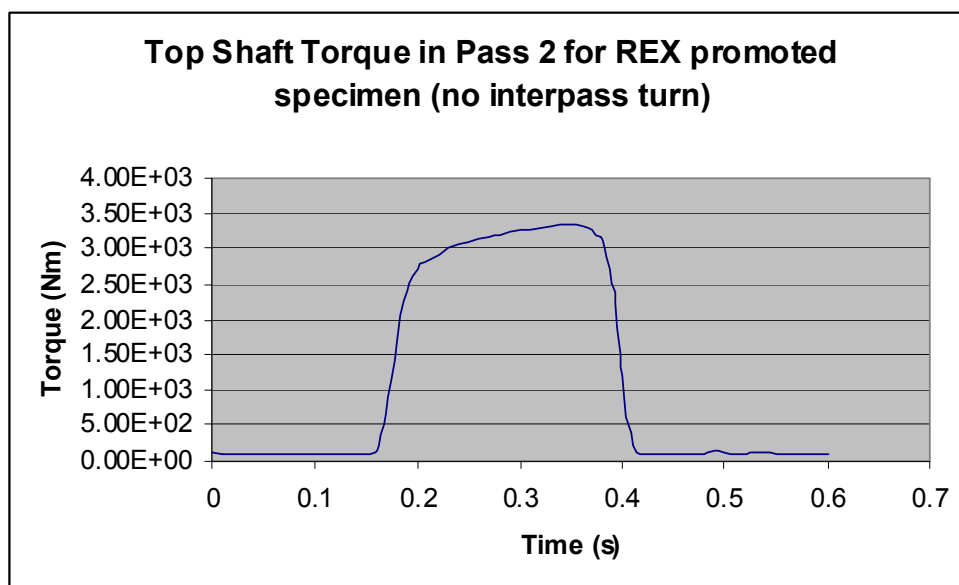


(b)

Fig. A1.51(a and b): Load and Torque Traces in Pass 2 for specimen between passes given single reheat to 950°C and turn through 90° in transverse – vertical plane

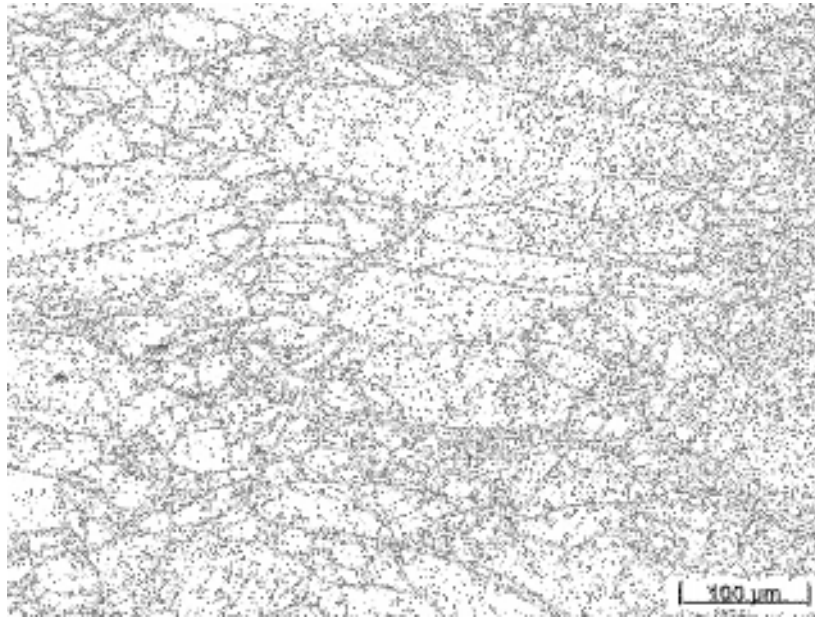


(a)

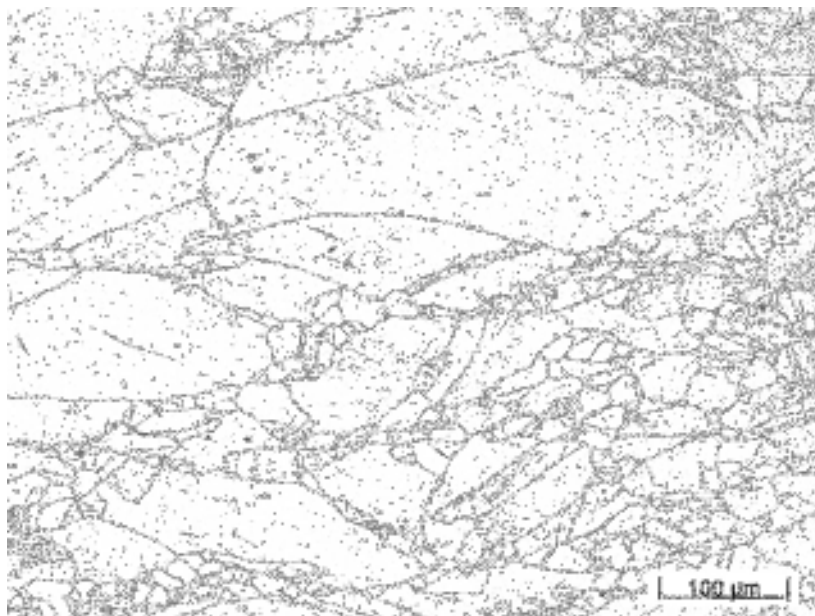


(b)

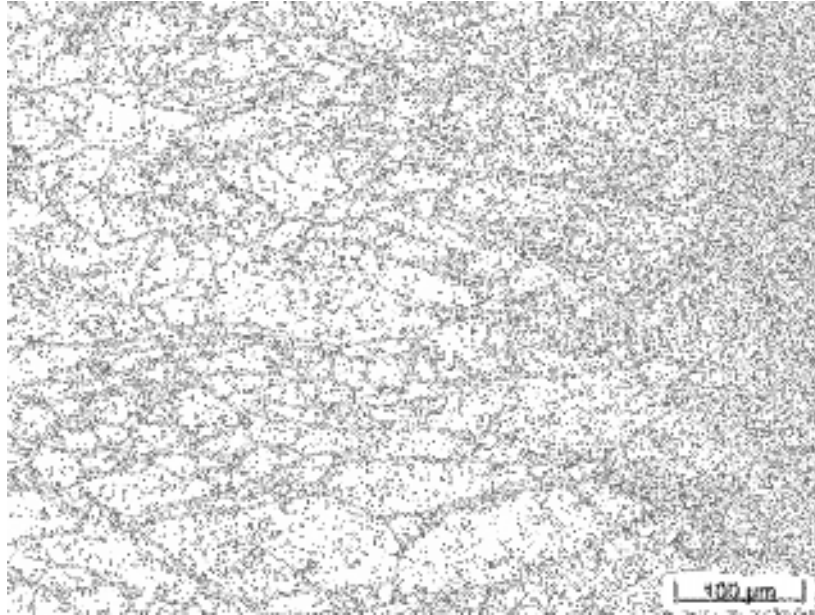
Fig. A1. 52(a and b): Load and Torque Traces for specimen in interpass given reheat to 1040°C- held for 60 s, followed by quench and further reheat to 950°C but no turn



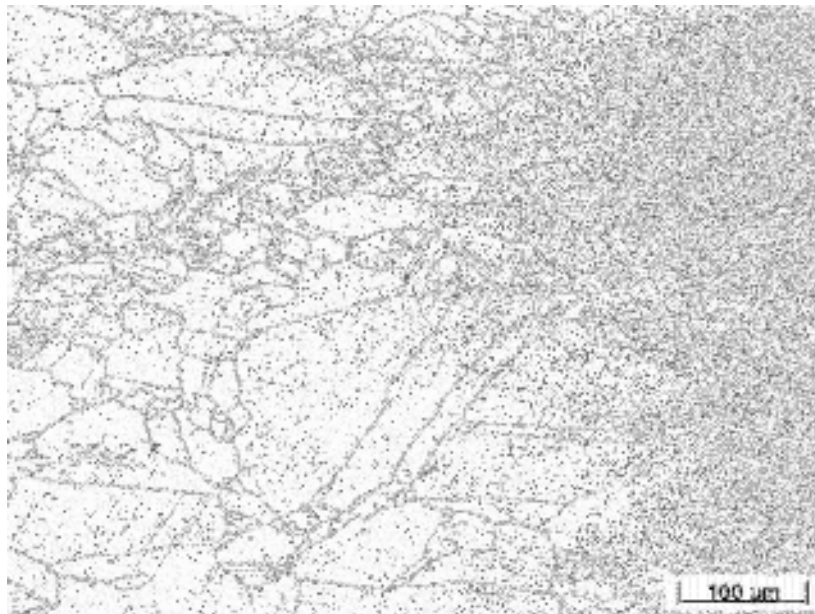
(a)



(b)

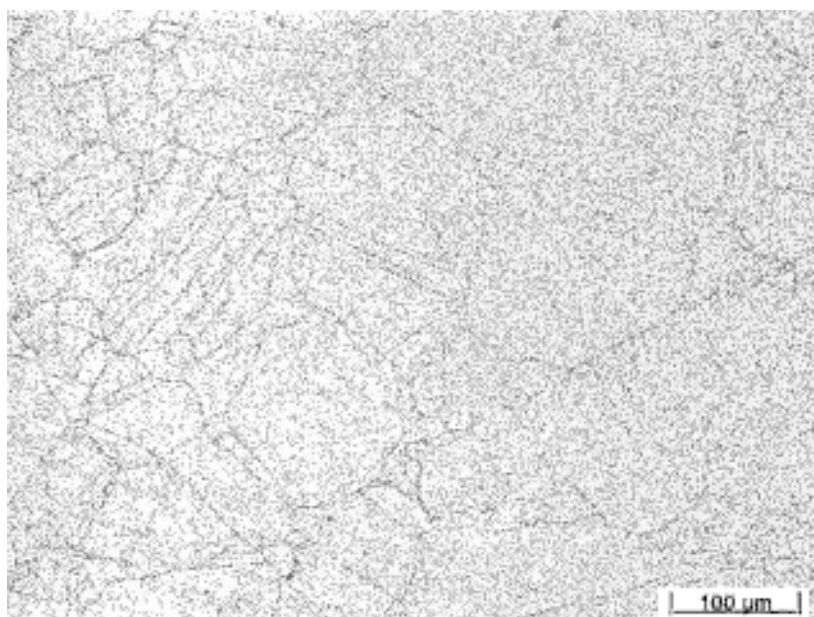


(c)

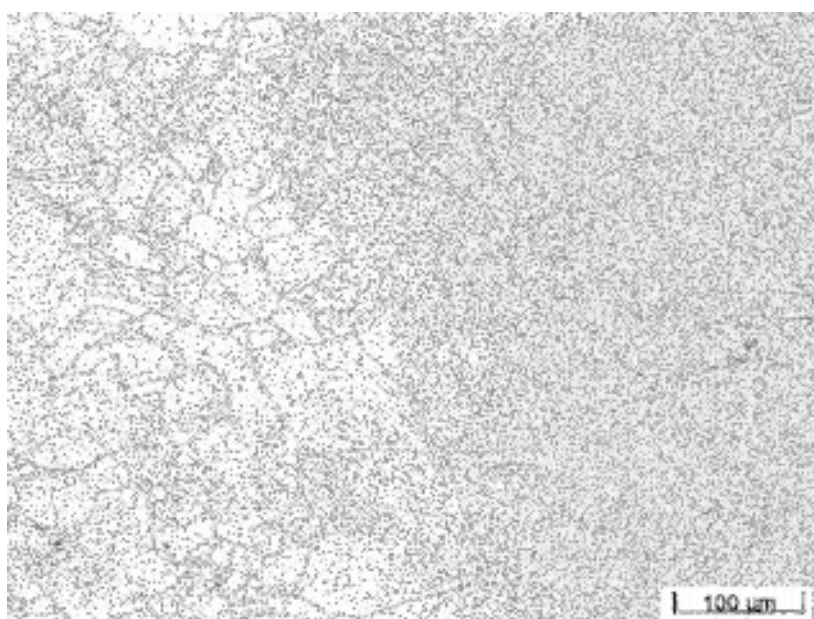


(d)

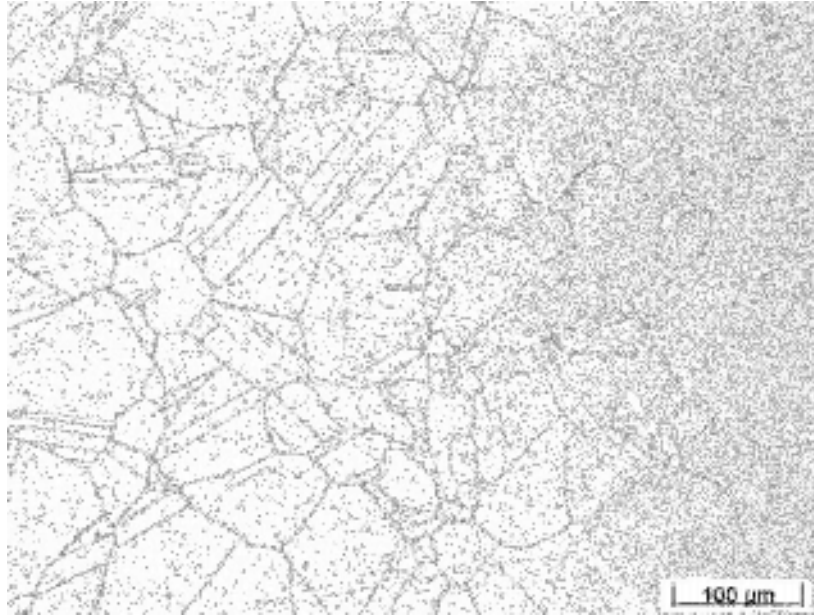
Fig. 53(a-d): Micrographs of Type 316 Stainless steel sample taken from bar 4R1F284A1 (a) , (b) held at 950°C for 10 s, resp. at 1/4depth and centre positions (c), (d) held for 60 s at 1/4depth and centre positions



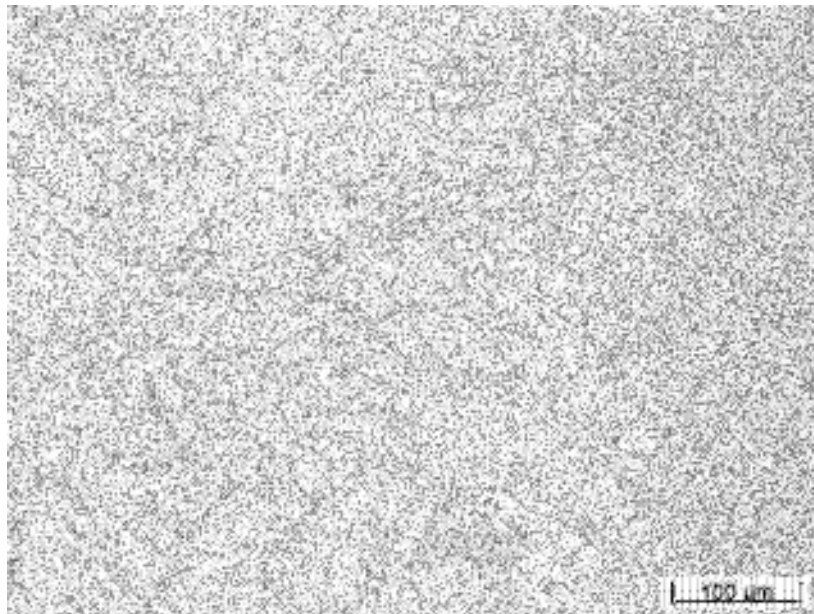
(a)



(b)

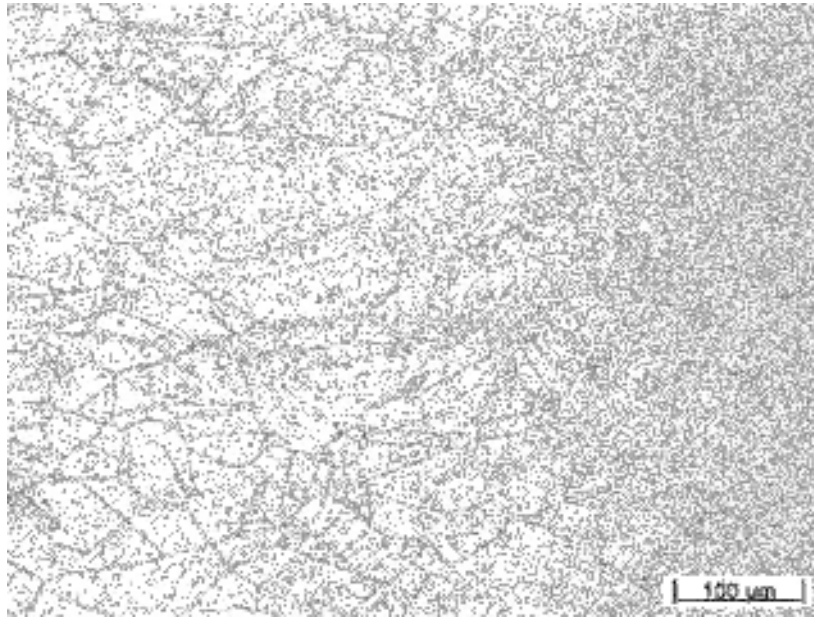


(c)

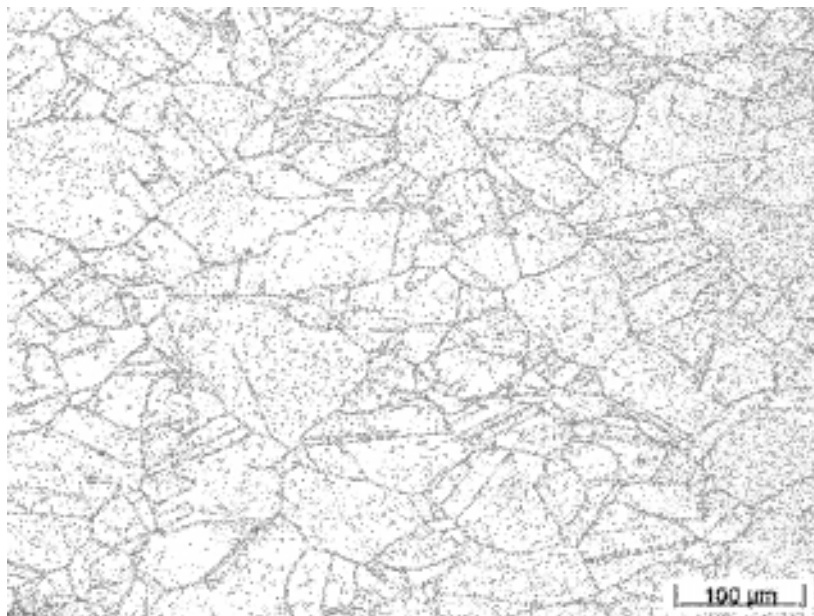


(d)

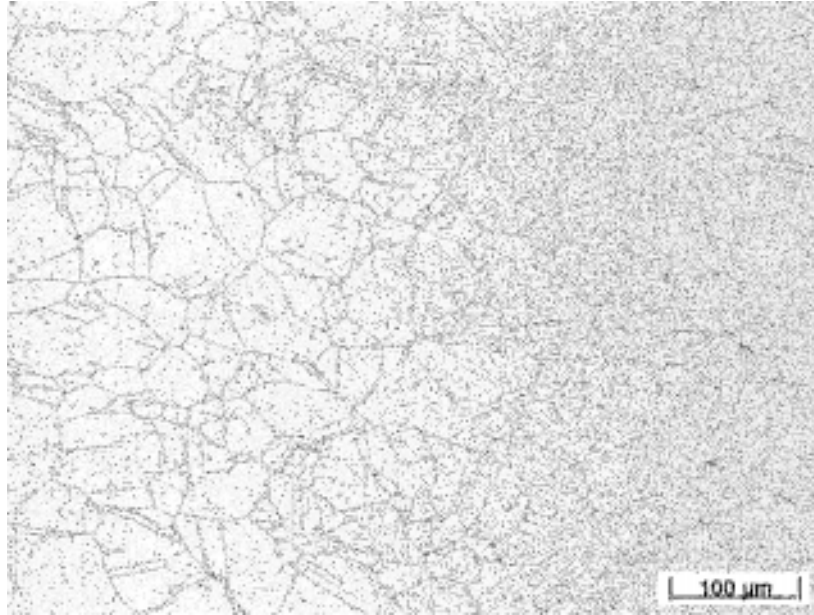
Fig. 54(a-d): Micrographs of Type 316 Stainless steel sample taken from bar 4R1F284A2 (a) , (b) held at 950°C for 10 s, resp. at 1/4depth and centre positions (c), (d) held for 60 s at 1/4depth and centre positions



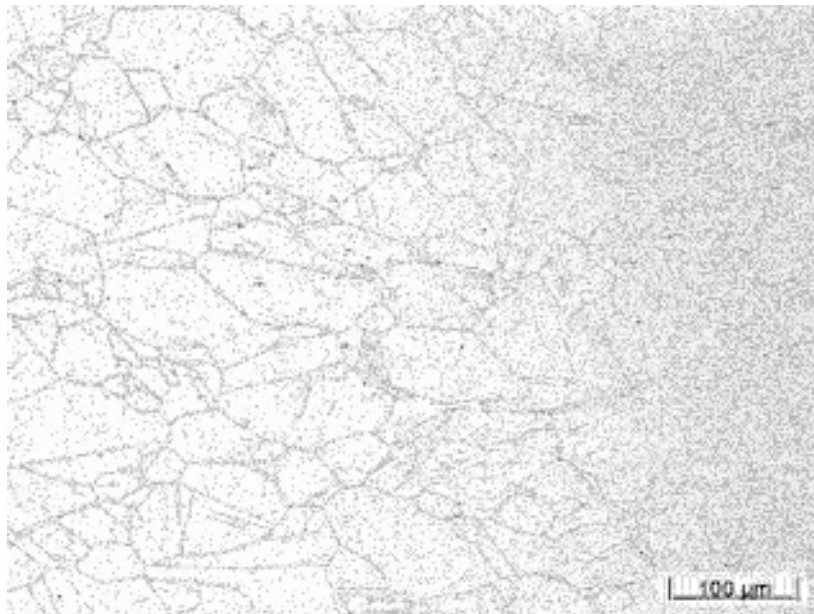
(a)



(b)

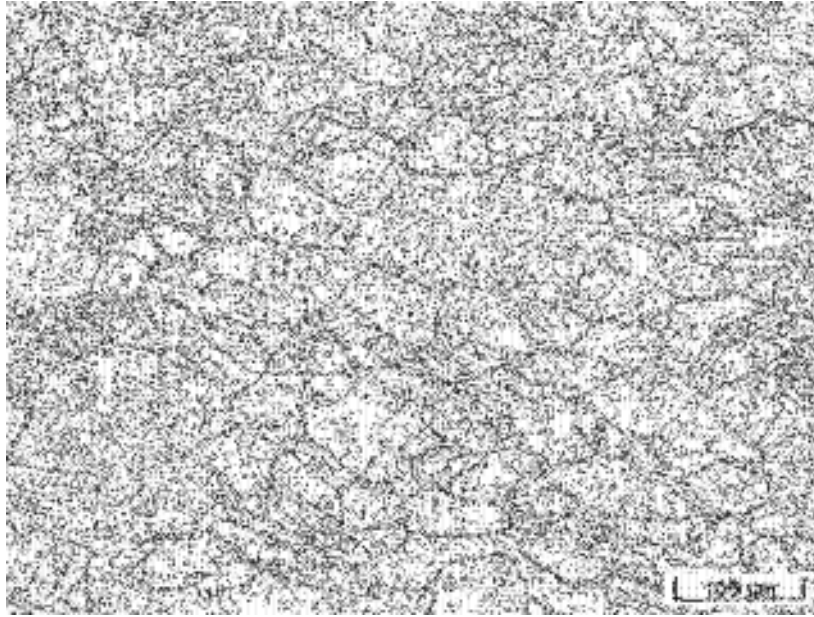


(c)

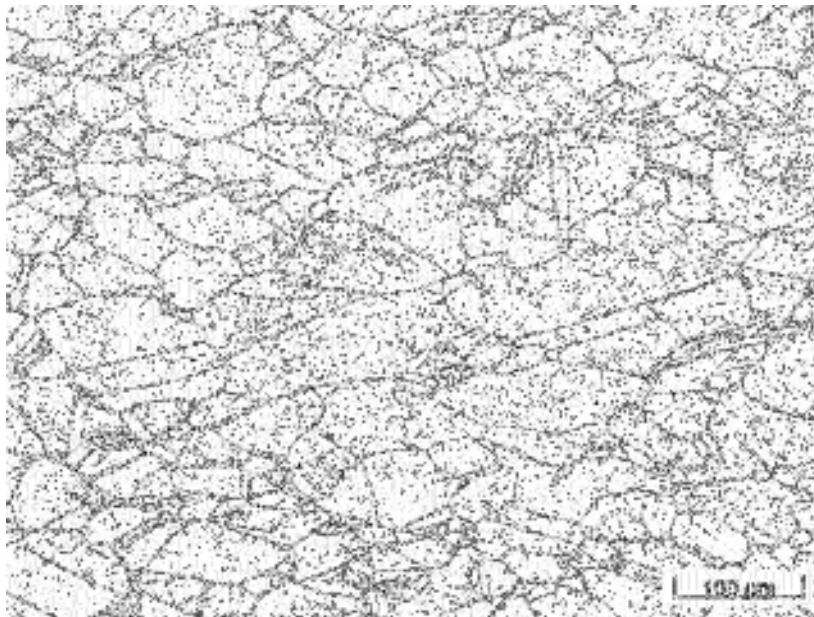


(d)

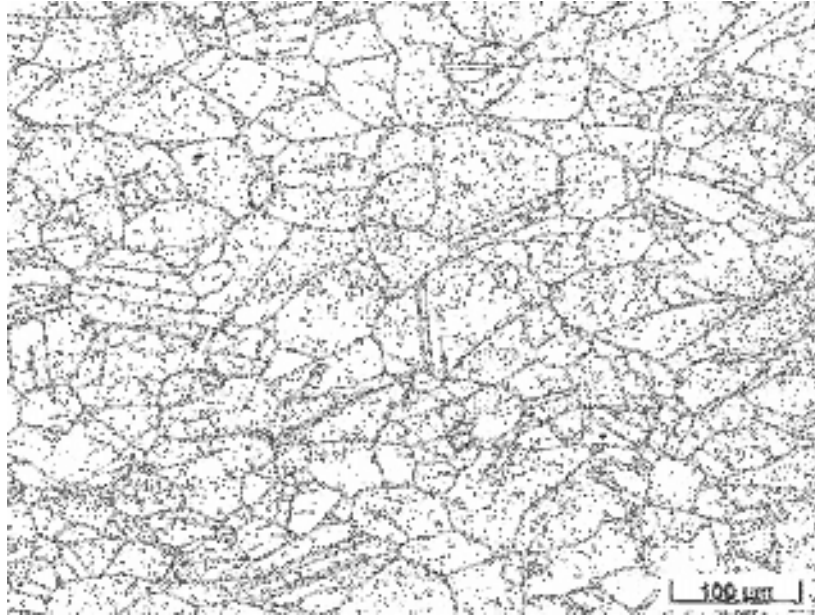
Fig. 55(a-d): Micrographs of Type 316 Stainless steel sample taken from bar 4R1F64A1 (a), (b) held at 950°C for 10 s, resp. at 1/4depth and centre positions (c), (d) held for 60 s at 1/4depth and centre positions



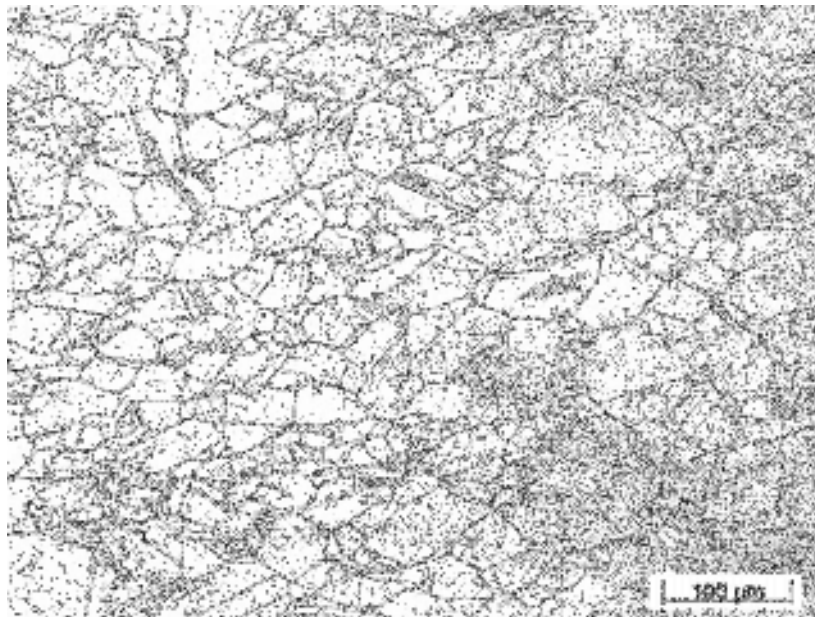
(a)



(b)



(c)



(d)

Fig. 56(a-d): Micrographs of Type 316 Stainless steel sample taken from bar 4R1F64A2 (a), (b) held at 950°C for 10 s, resp. at 1/4depth and centre positions (c), (d) held for 60 s at 1/4depth and centre positions

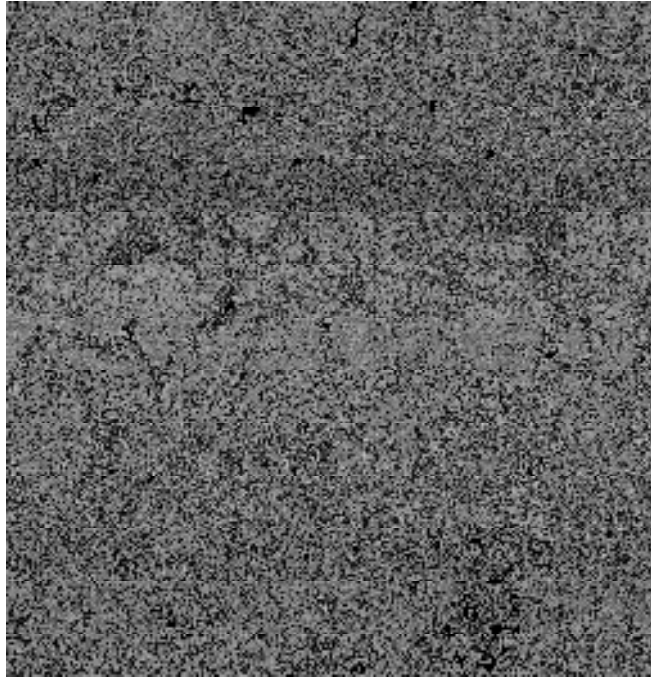


Fig. A1.57. Image Quality map obtained by EBSD for sample 4R1F64A1B (with extra reheat)
(400 microns x 413 microns)

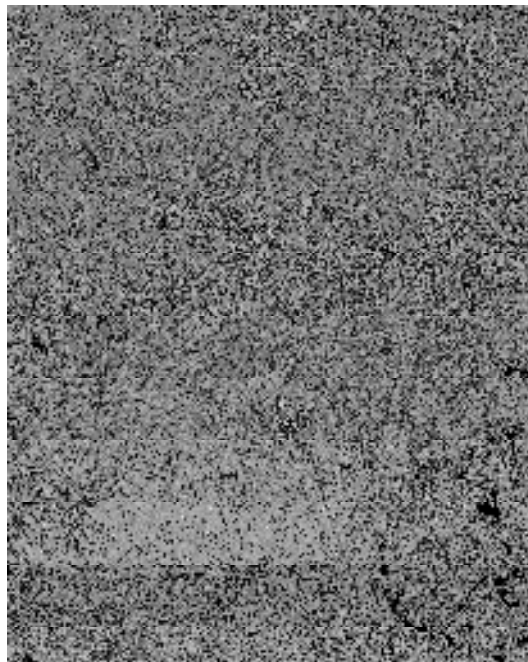
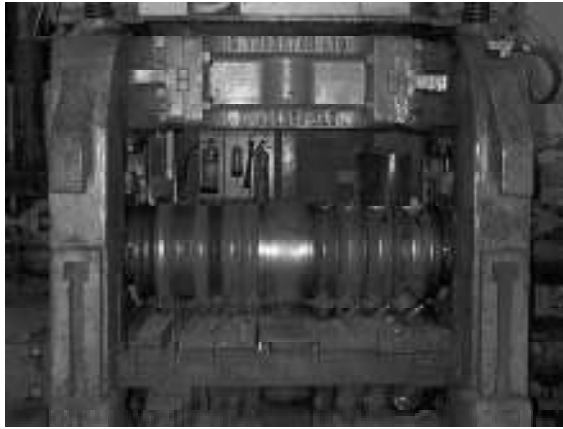


Figure A1.58: Image Quality map obtained by EBSD
for sample 4R1F284A1B (without extra reheat)(400 microns x 500 microns)



(a) 3 High Mill



(b) Furnace

Fig. A1.59(a and b): Facilities for plate reversal rolling trials

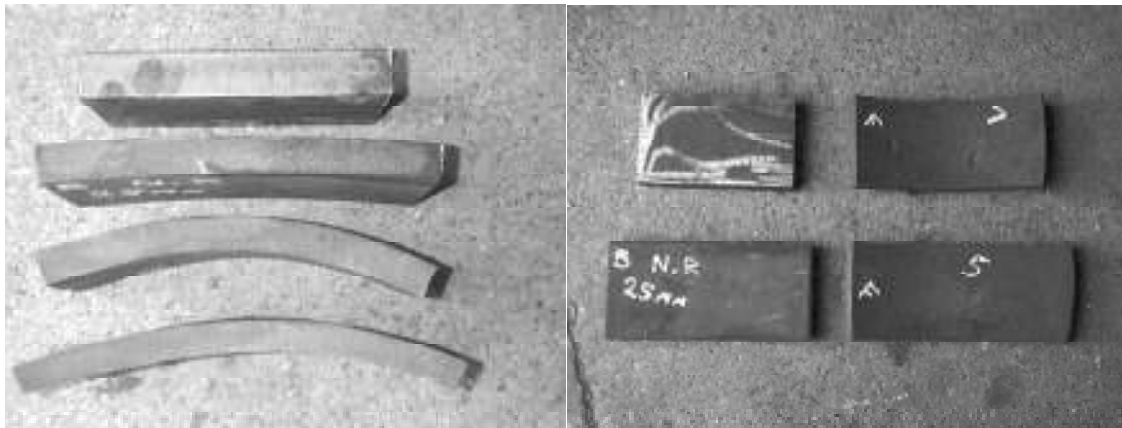
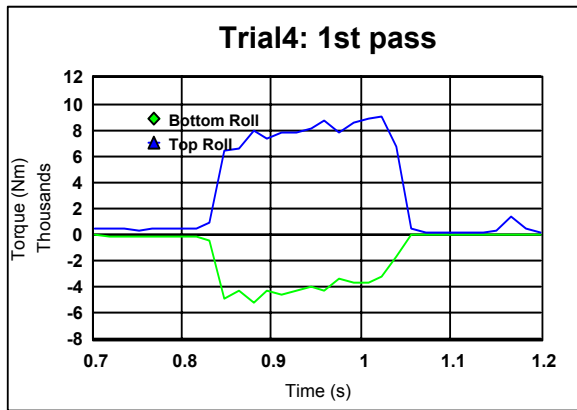
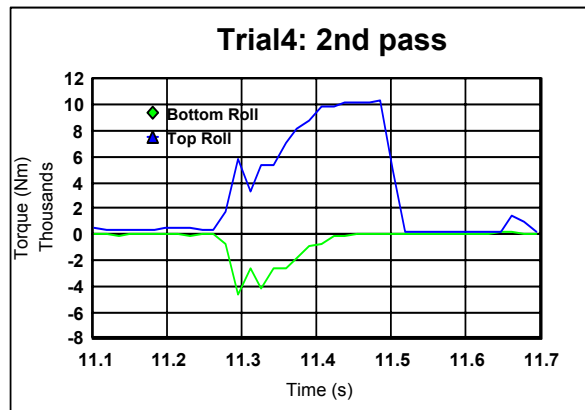


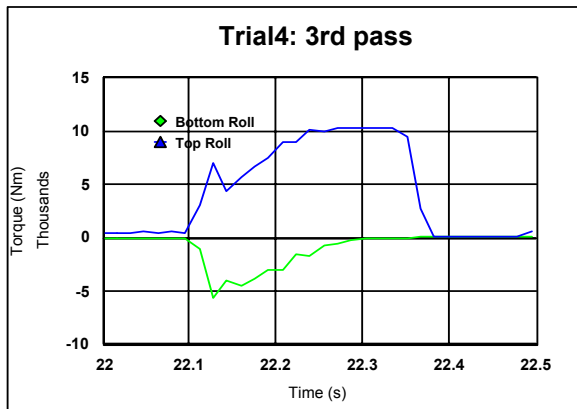
Fig. A1.60: Rolled profiles of the samples



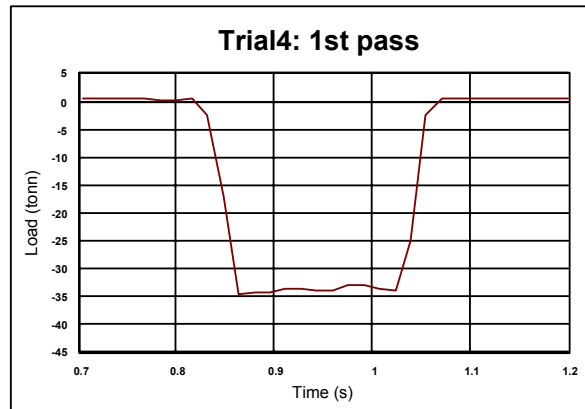
(a) Torque in the 1st pass



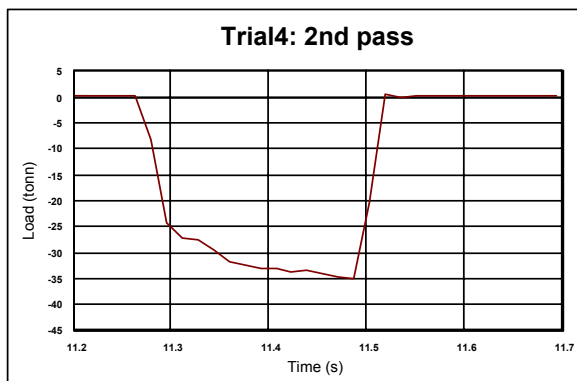
(b) Torque in the 2nd pass



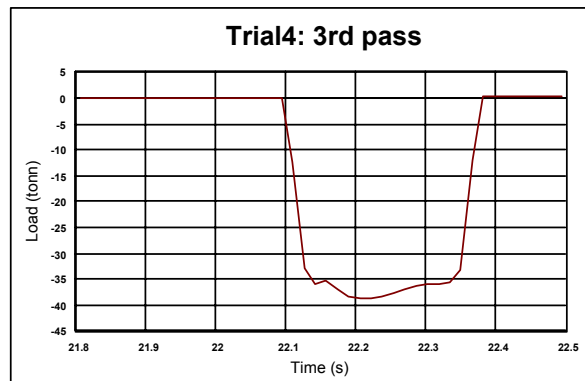
(c) Torque in the 3rd pass



(d) Load in the 1st pass

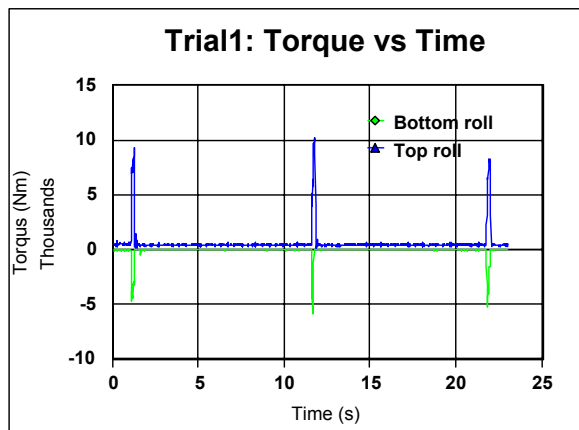


(e) Load in the 2nd pass

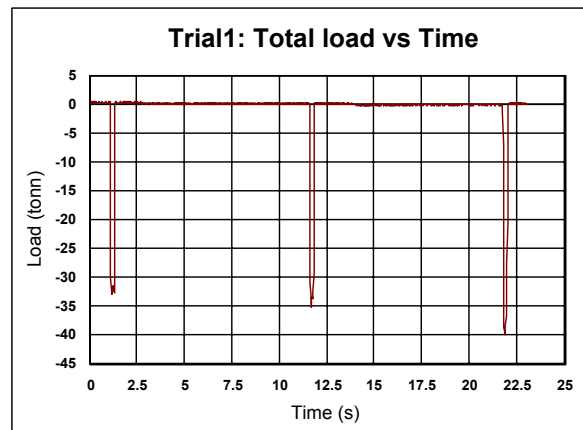


(f) Load in the 3rd pass

Fig. A1.61(a-f): Load and Torque in Trial 4



(a) Torque



(b) Load

Fig. A1.62(a and b): Load and Torque in Trial 1

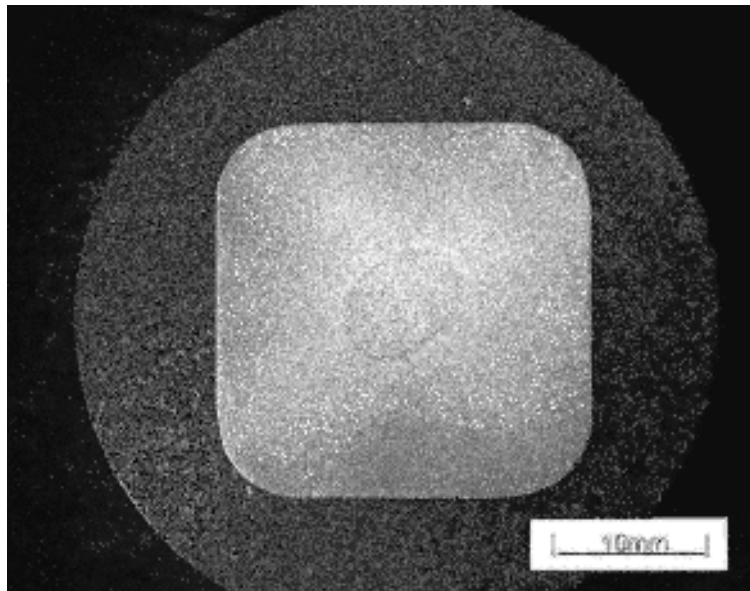
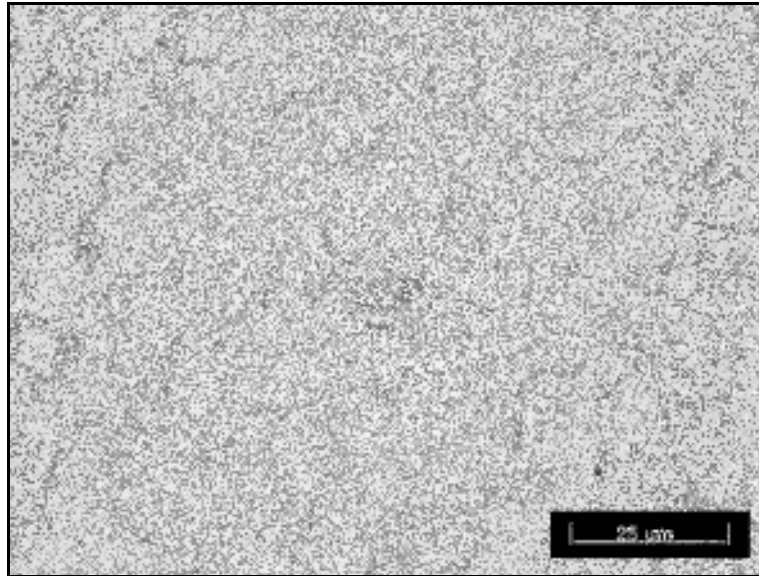
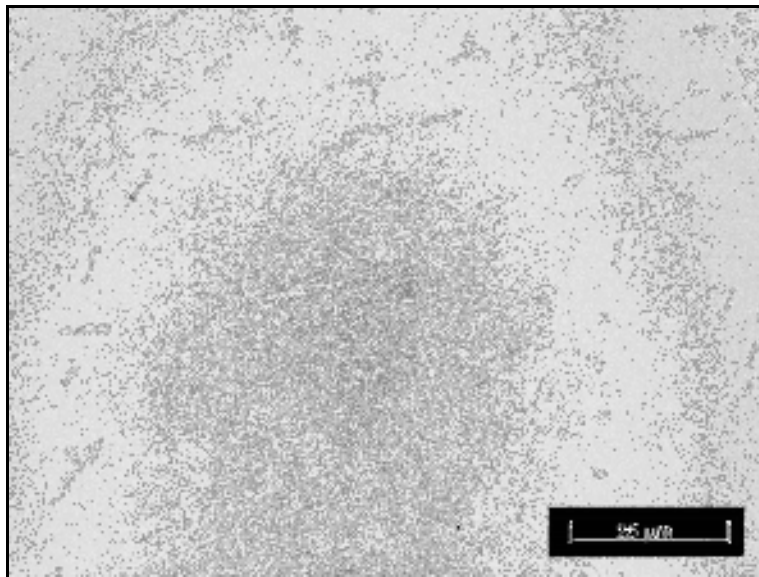


Fig. A1.63: Etch of as-supplied Type 316 stainless steel specimen contrasting central core and outer layers (Macro JobK03M00208)



x 1000

Fig. A1.64: Microstructure of Type 316 stainless steel specimen at quarter depth position



x 1000

Fig. A1.65: Microstructure of Type 316 stainless steel specimen showing second phase of delta ferrite

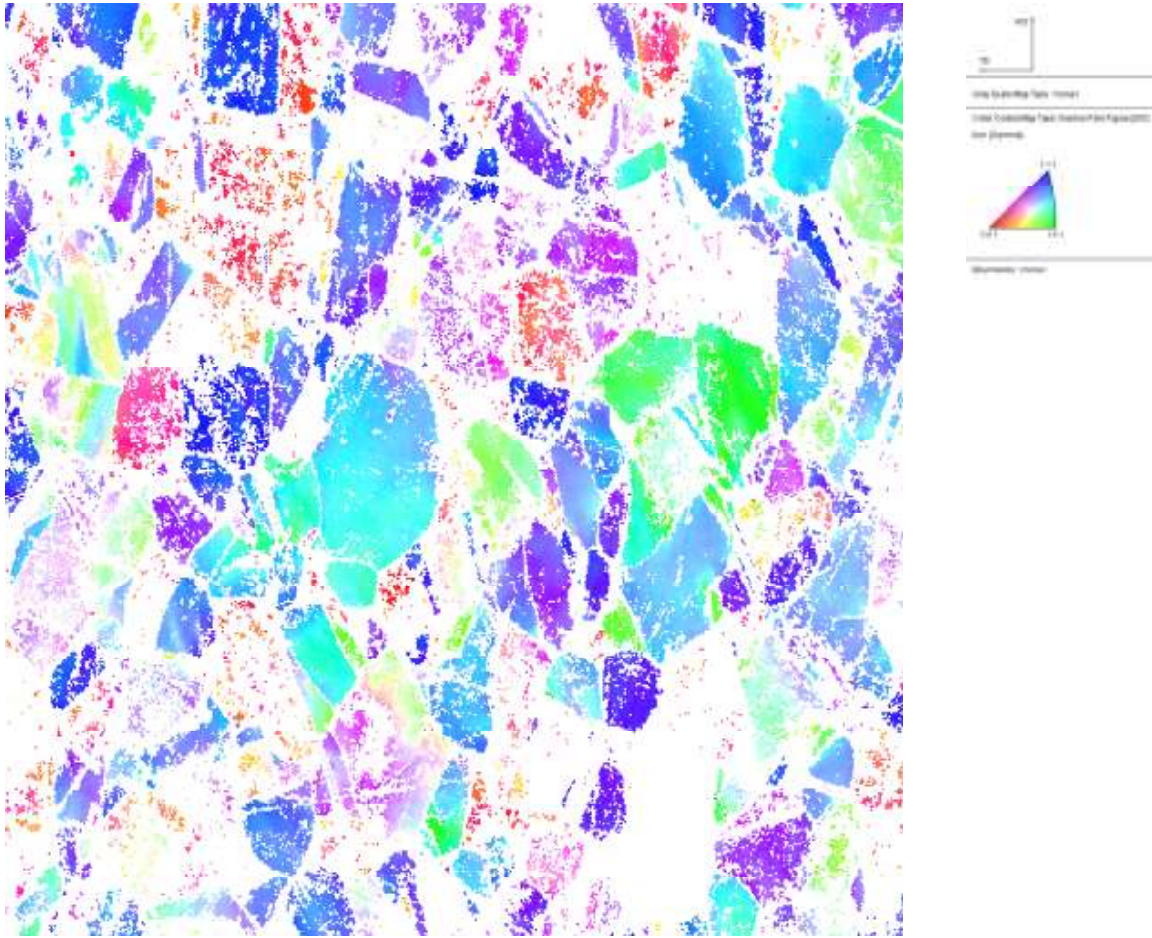


Fig. A1.66: IPF map (and key) for sample 4R1F64A1B (400 microns x 413 microns)

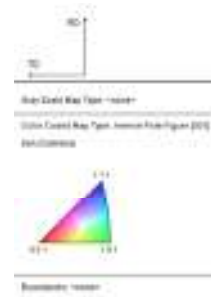
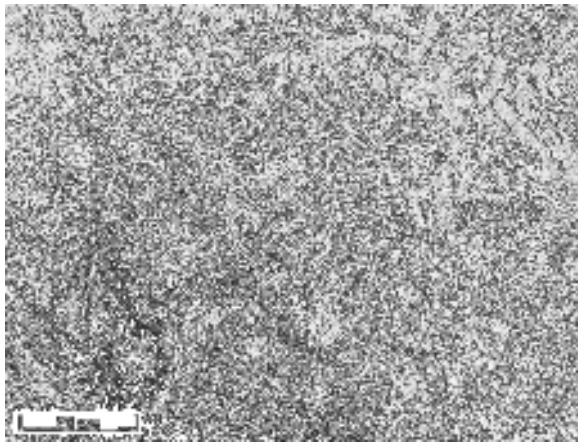
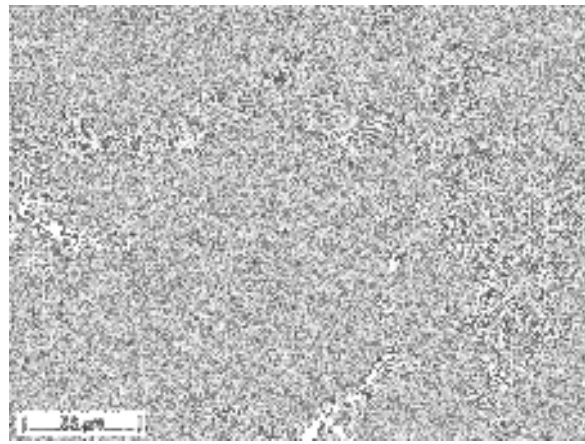


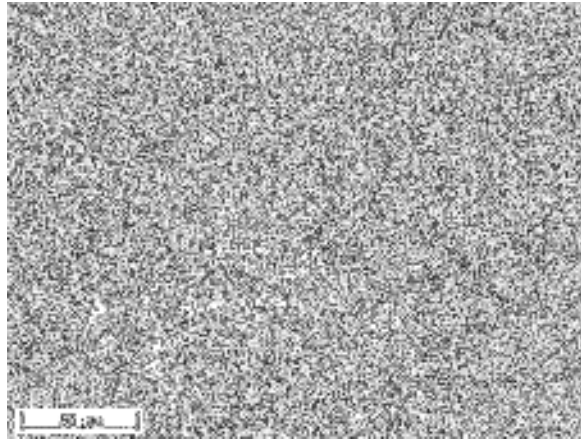
Fig. A1.67: IPF map (and key) for sample 4R1F284A1B (400 microns x 500 microns)



(a) Centre

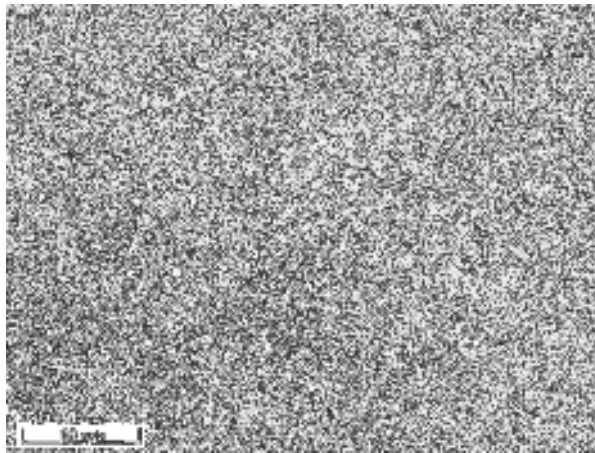


(b) Surface 1

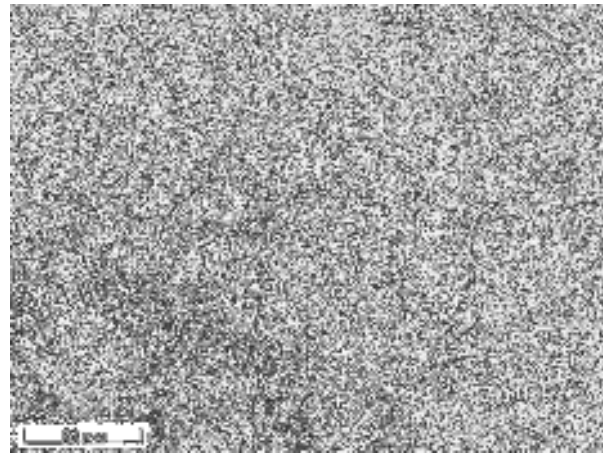


(c) Surface 2

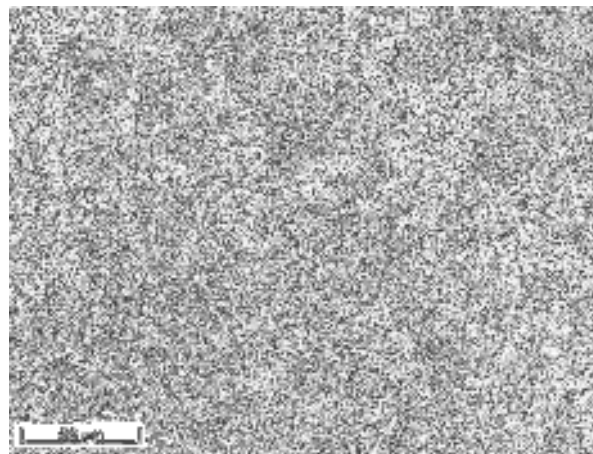
Fig. A1.68(a-c): Metallography of trial 1 in forward–reverse rolling



(a) Centre

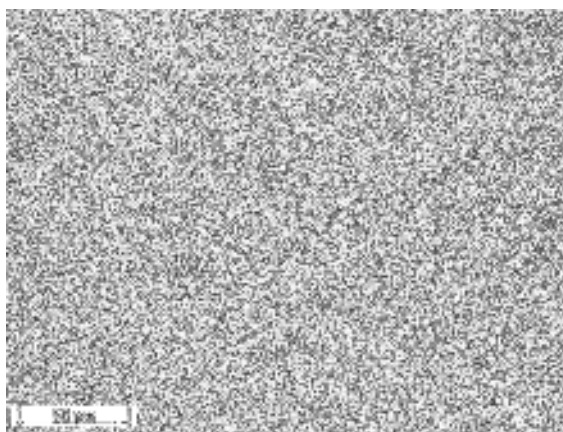


(b) Surface 1

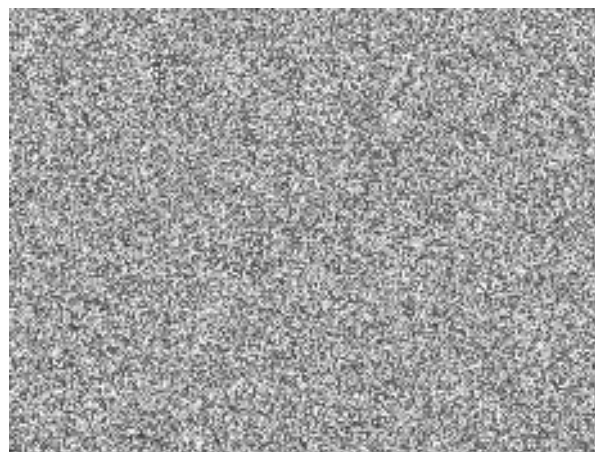


(c) Surface 2

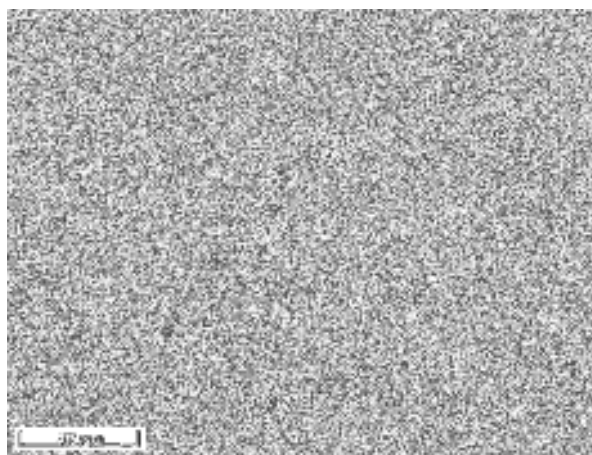
Fig. A1.69(a-c): Metallography of Trial 4 in forward-reverse rolling



(a) Centre

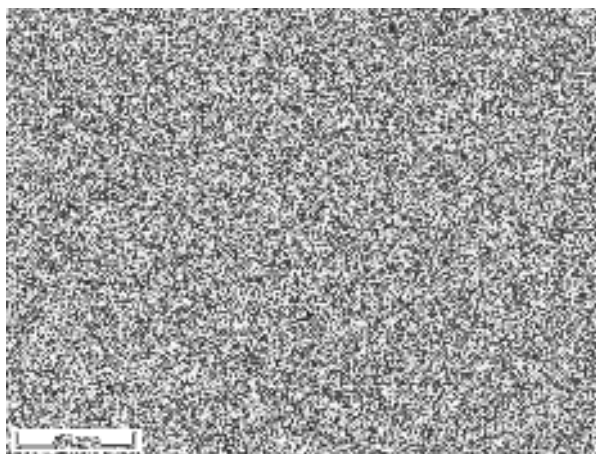


(b) Surface 1

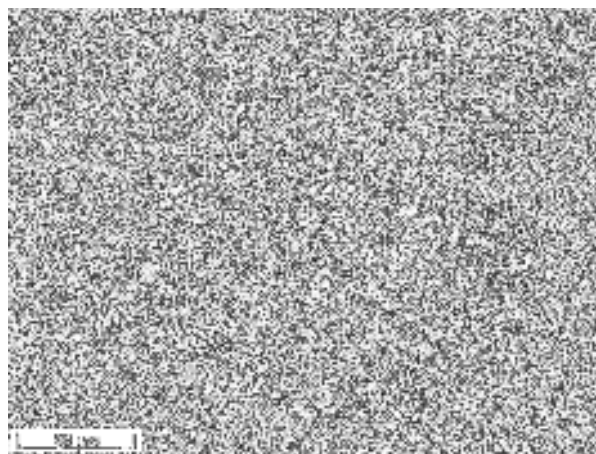


(c) Surface 2

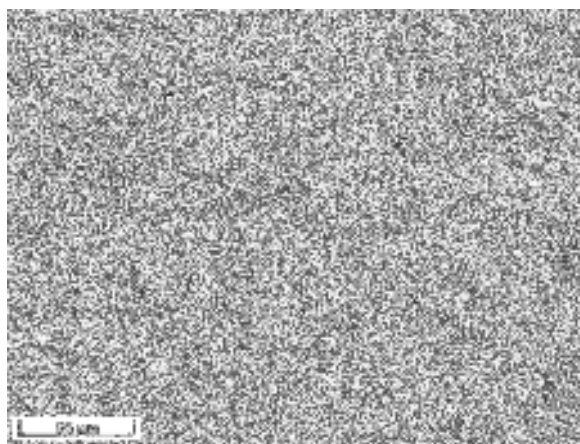
Fig. A1.70: Metallography of Trial 7 in forward-reverse rolling



(a) Centre

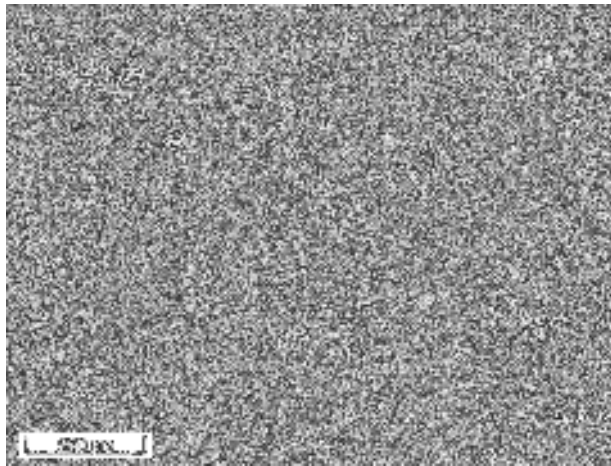


(b) Surface 1

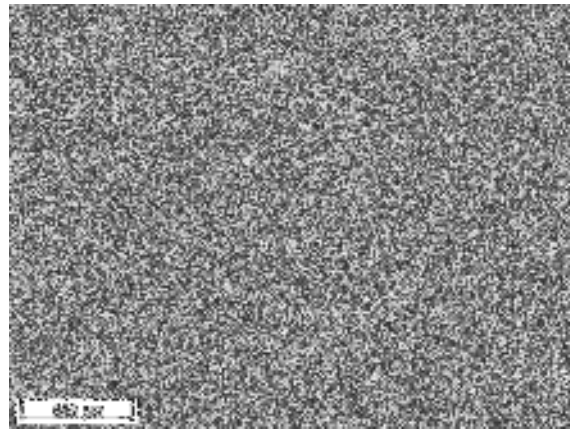


(c) Surface 2

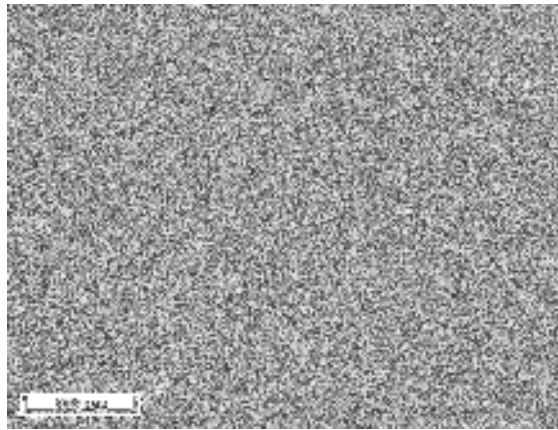
Fig. A1.71(a-c): Metallography of Trial 8 in forward-reverse rolling



(a) Centre

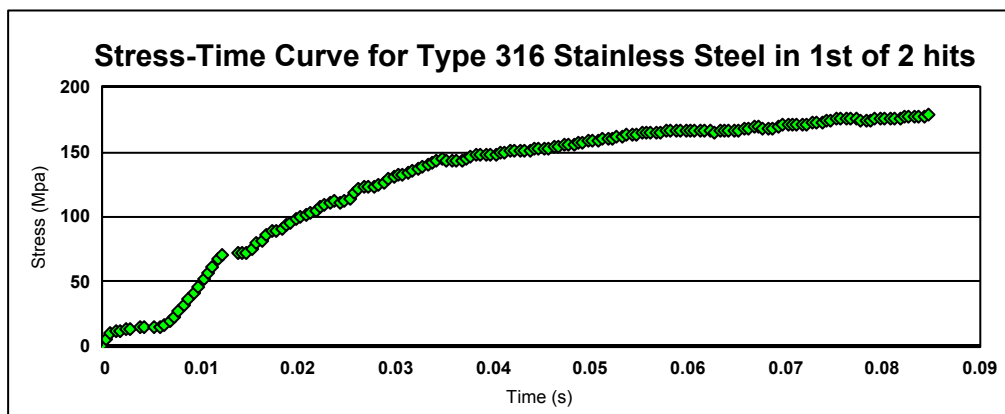


(b) Surface 1

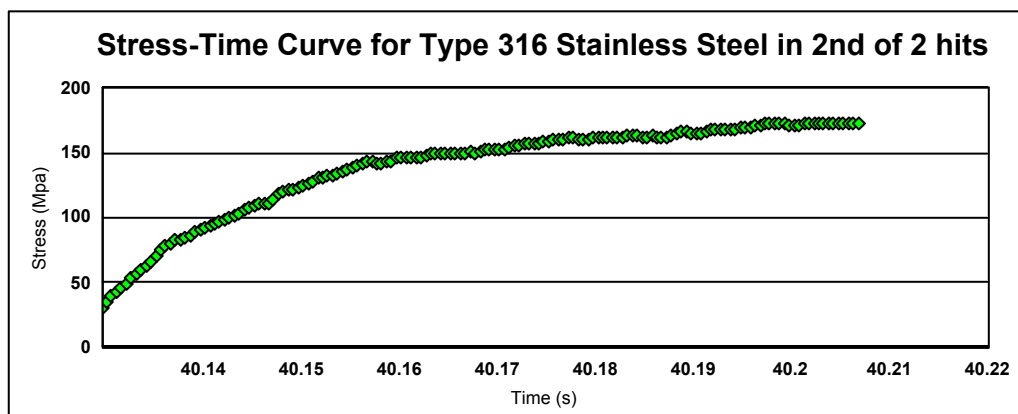


(c) Surface 2

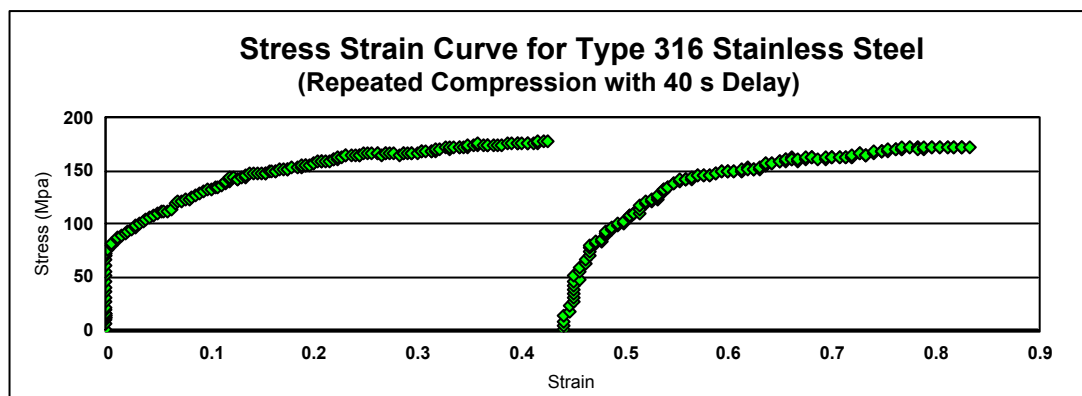
Fig. A1.72(a-c): Metallography following repolishing etc. of Trial 7 (exhibiting martensite) in forward-reverse rolling



(a)

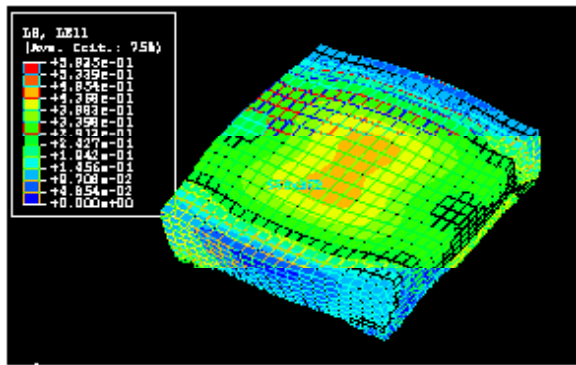


(b)

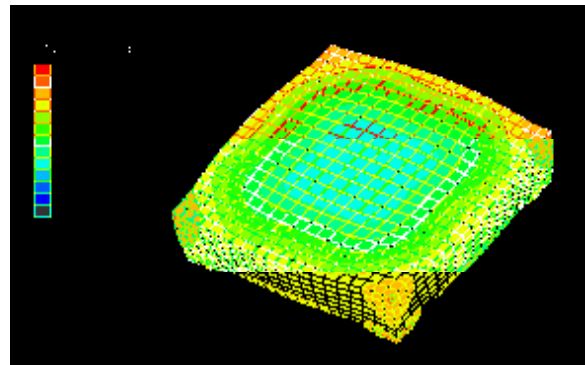


(c)

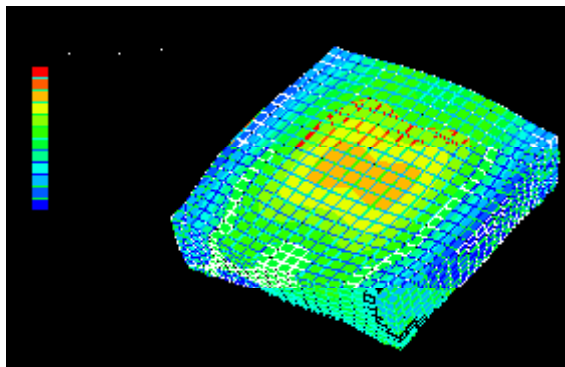
Fig. A1.73(a-c): Curves showing, for each hit of double compression test, the stress-time and stress-strain



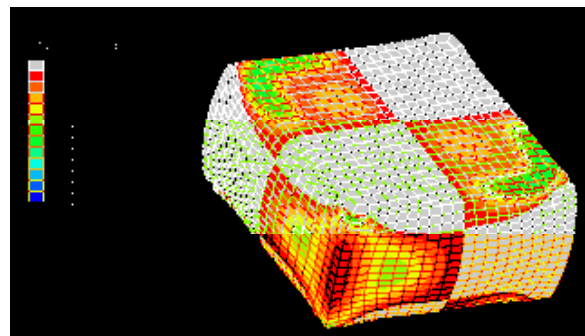
(a) Longitudinal normal



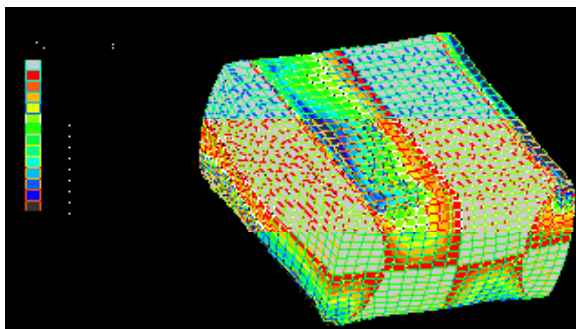
(b) Vertical normal component



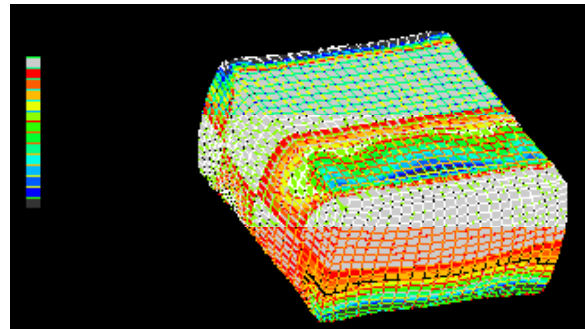
(c) Transverse normal component



(d) Longitudinal-transverse shear component



(e) Transverse-vertical shear component



(f) Longitudinal-vertical shear component

Fig. A1.74(a-f): Predicted components of logarithmic strain in 10 mm cube of type 316 stainless steel after 40% reduction in Gleeble machine at 1000°C (predictions a-c in mid-horizontal plane)

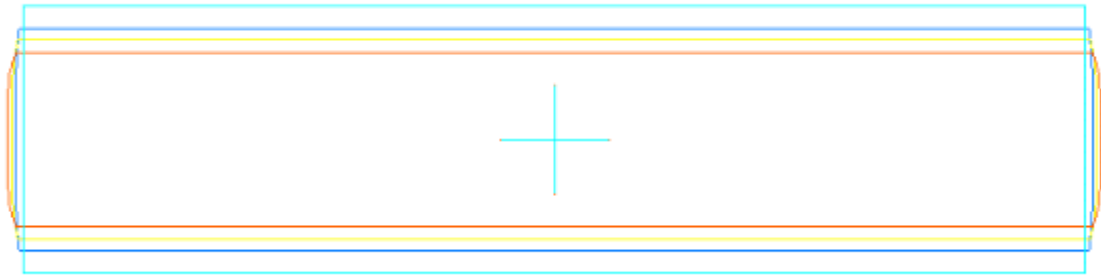


Fig. A1.75: Transverse profile predicted by FEM for 3 passes of trial
(spread values given in Table A1.2)

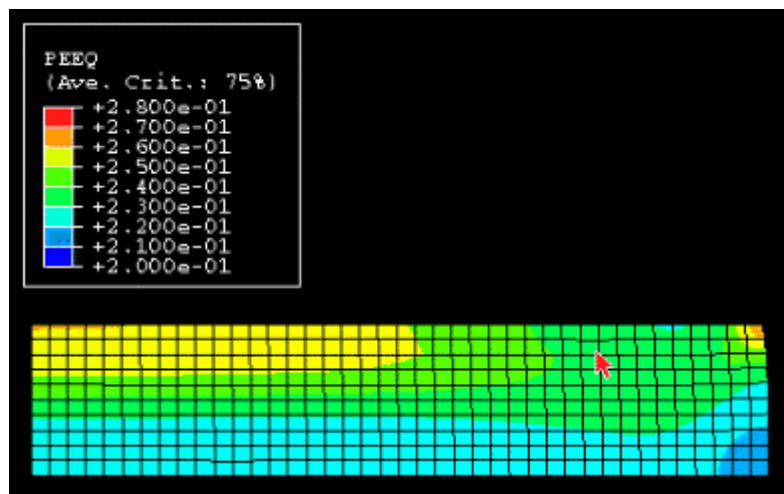


Fig. A1.76: Plate reversal rolling: predicted equivalent plastic strain distribution in quarter
section of plate in Pass 1 of Trial 1

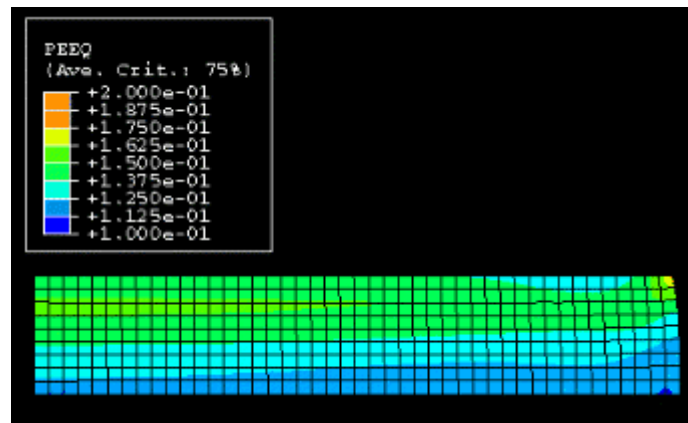


Fig. A1.77: Plate reversal rolling: predicted equivalent plastic strain distribution in quarter
section of plate in Pass 2 of Trial 1

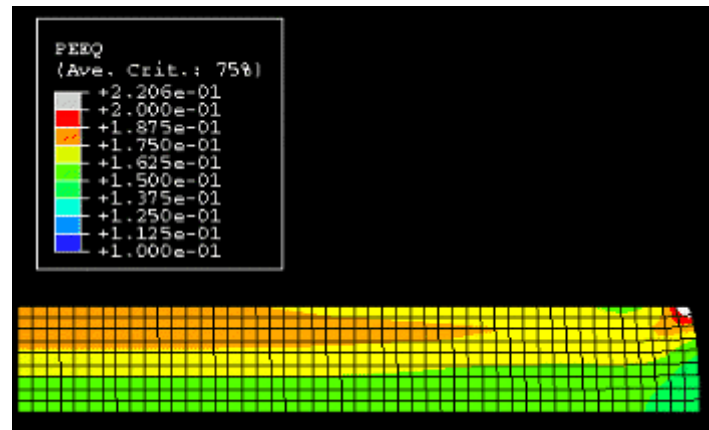
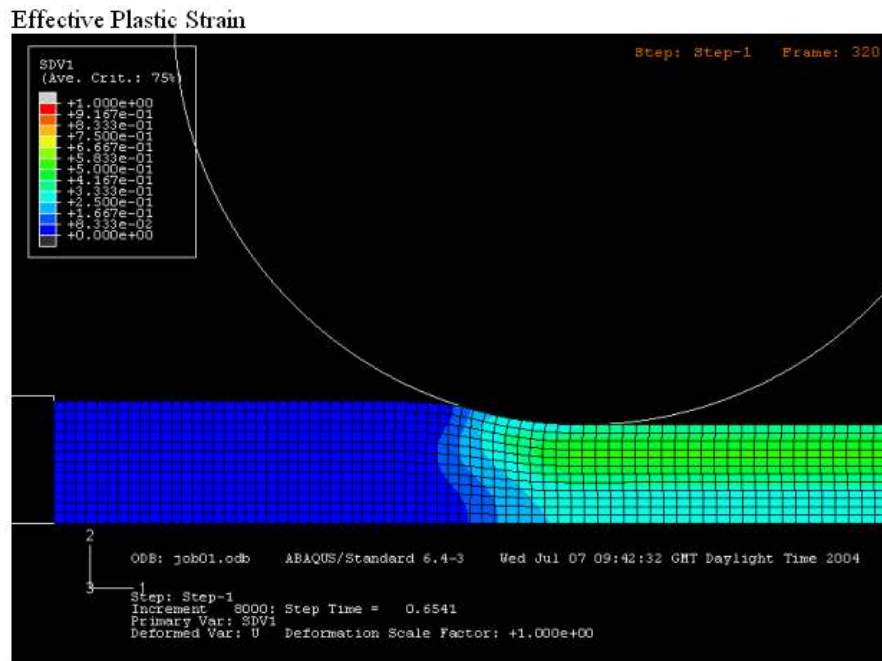
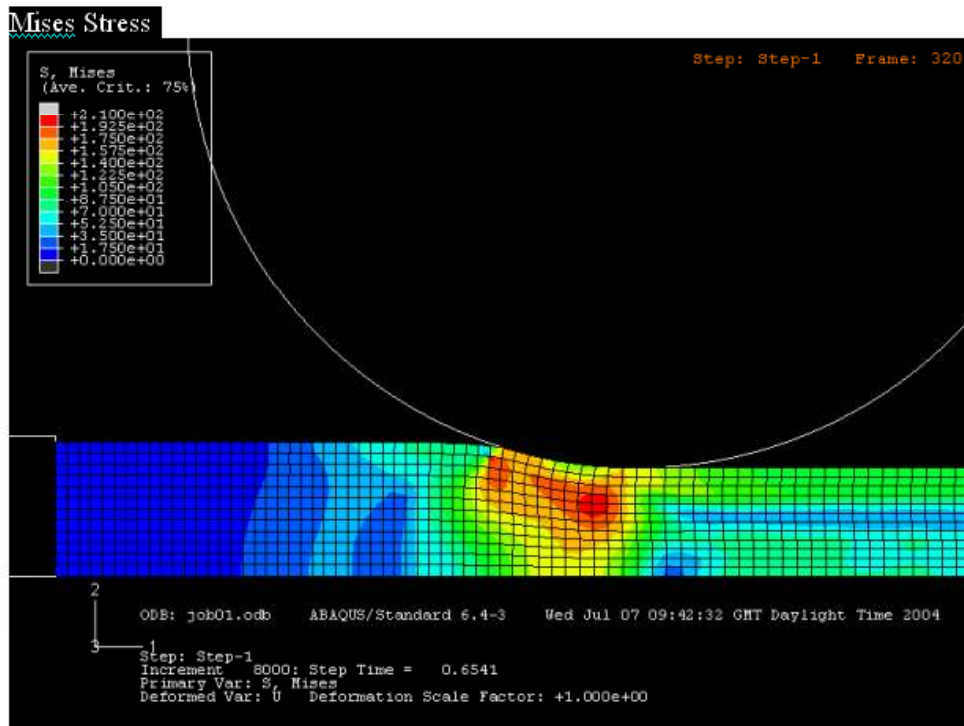


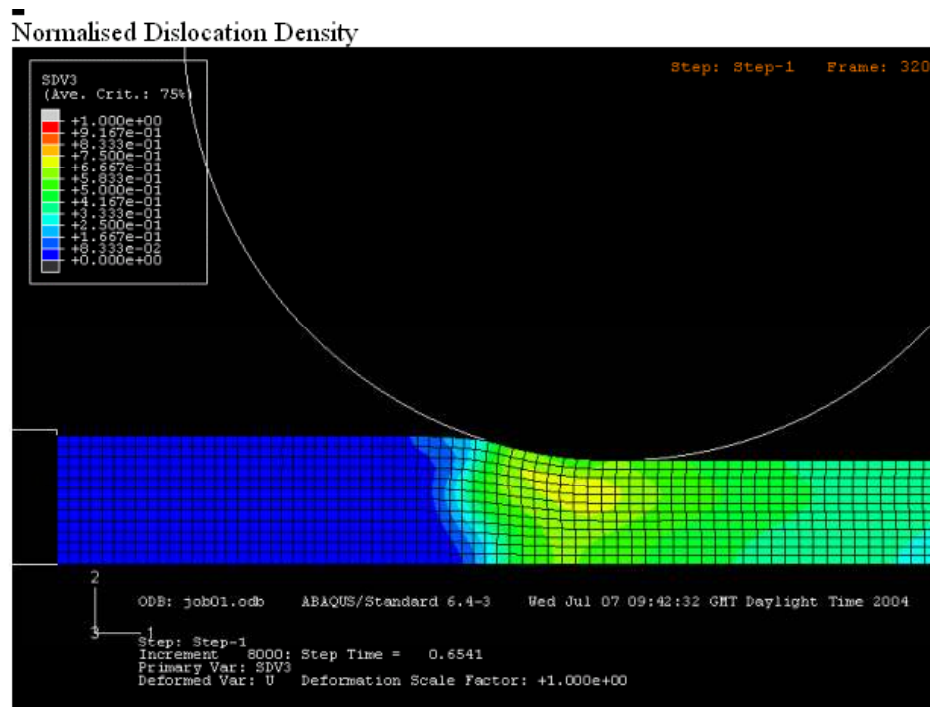
Fig. A1.78: Plate reversal rolling: predicted equivalent plastic strain distribution in quarter section of plate in Pass 3 of Trial 1



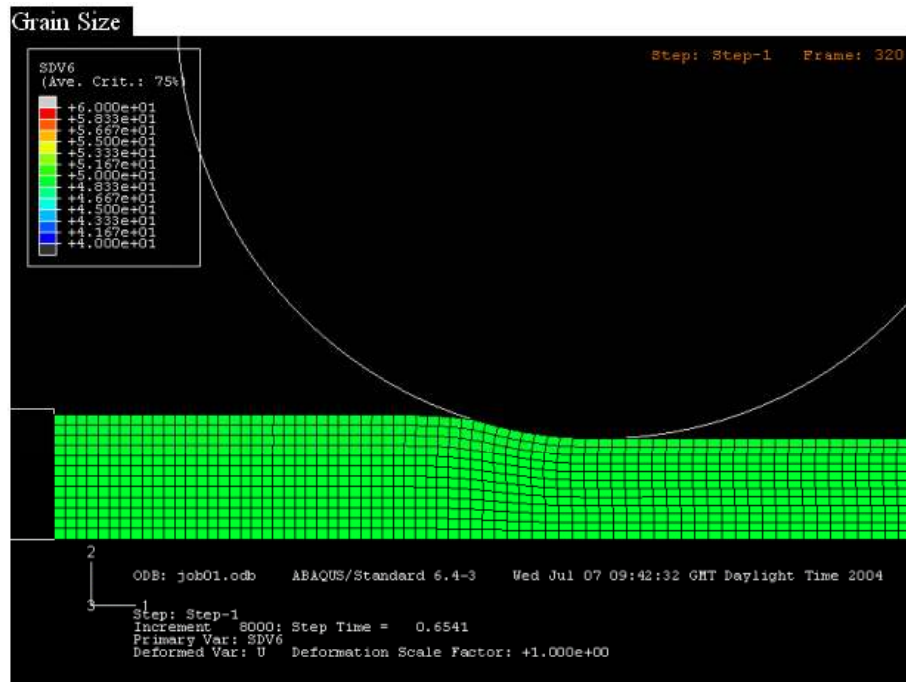
(a)



(b)

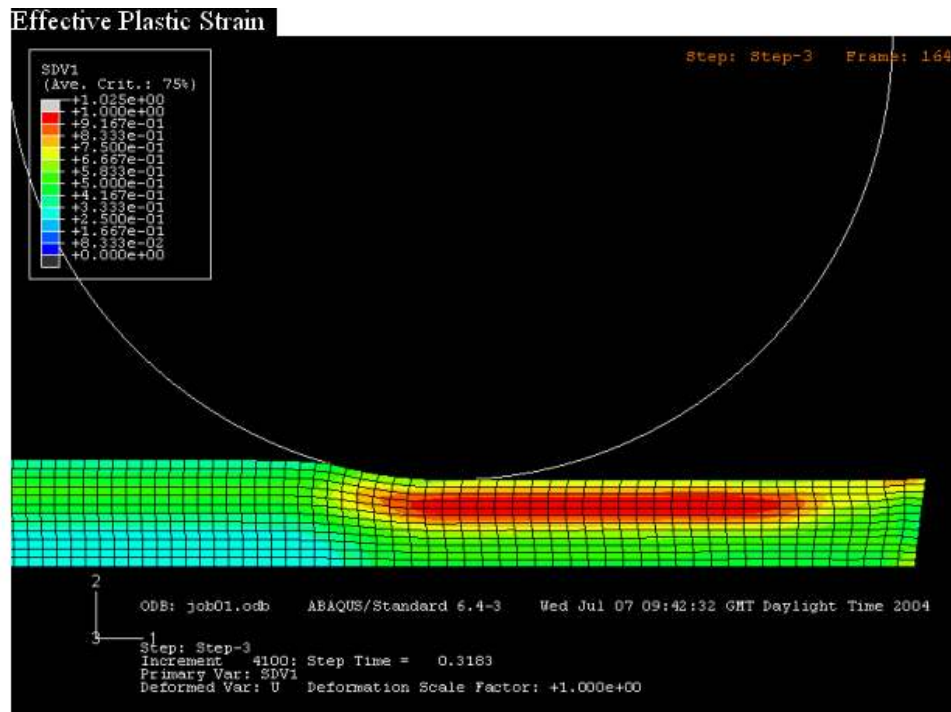


(c)

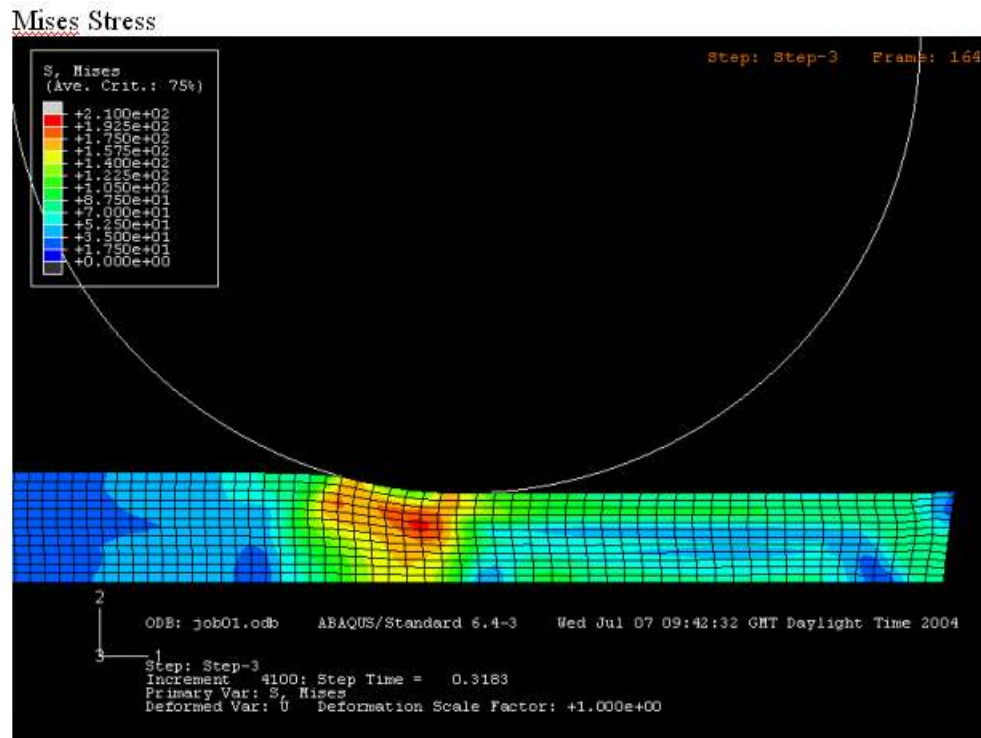


(d)

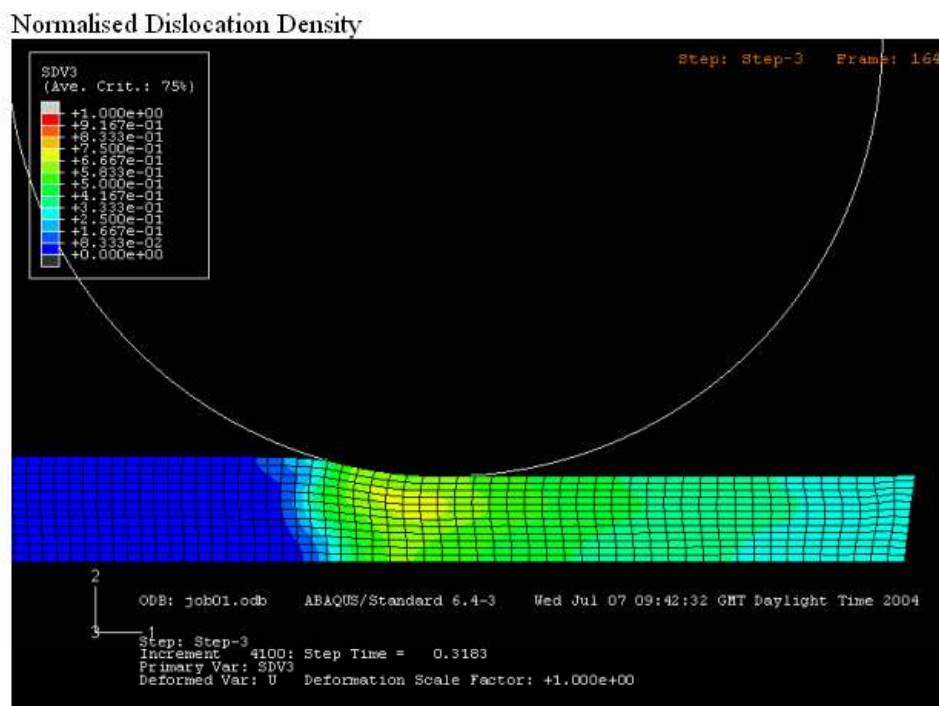
Fig. A1.79(a-d): Predictions, for pass 1 of H-H rolling schedule, of PEEQ strain, Mises stress, dislocation density and grain size



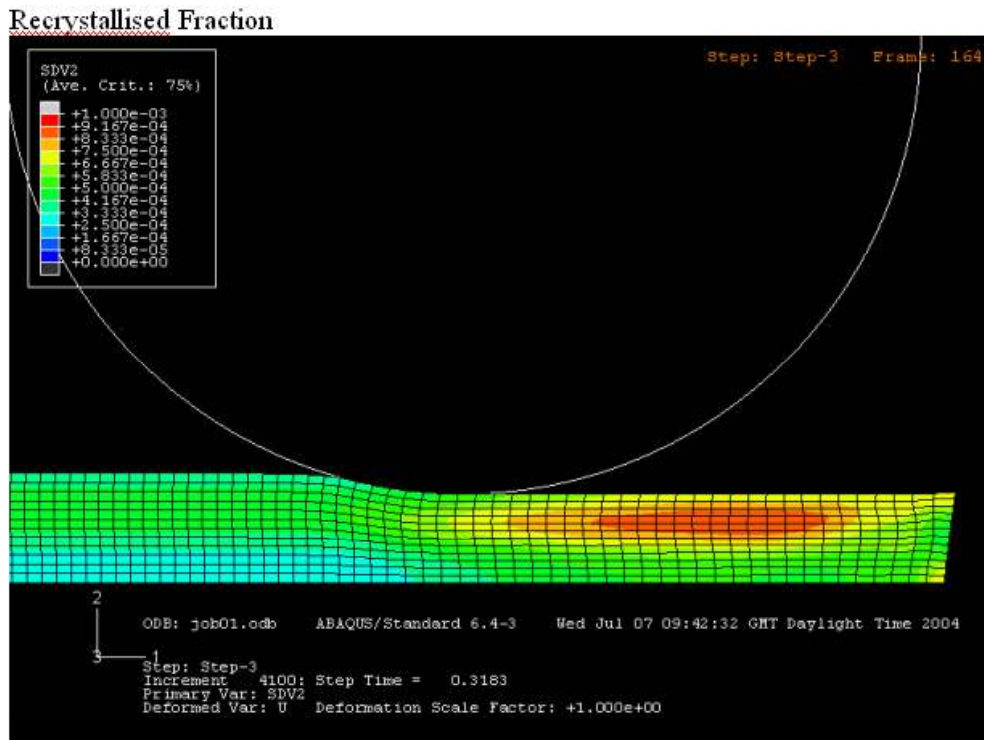
(a)



(b)



(c)



(d)

Fig. A1.80(a-d): Predictions, for pass 1 of H-H rolling schedule, of PEEQ strain, Mises stress, dislocation density and fraction recrystallised

Appendix 2

Detailed Technical Report by CEIT Authors: D Jorge-Badiola and I Gutierrez

Summary

The work carried out by CEIT in the present project has focused on the following topics:

- Performing hot torsion tests in order to obtain the flow curves under monotonic and strain reversal conditions for a set of steels: a Nb microalloyed steel, a medium carbon steel and a duplex stainless steel. These steels were provided by other partners.
- Modelling the hot working flow curves for the Nb microalloyed steel for monotonic and strain reversal conditions.
- Investigating the effect of the reversion of the strain on the recrystallisation for the Nb microalloyed and the medium carbon steels.
- Producing strain reversal cycles under pre-determined conditions in order to simulate the reversions of the shear strain taking place under the rolling rolls.
- Performing hot torsion reversal tests and sequences on polished and marked as-cast and wrought duplex stainless steel specimens in order to investigate the effect of the reversal on the γ - δ interface and to connect it with damage formation.

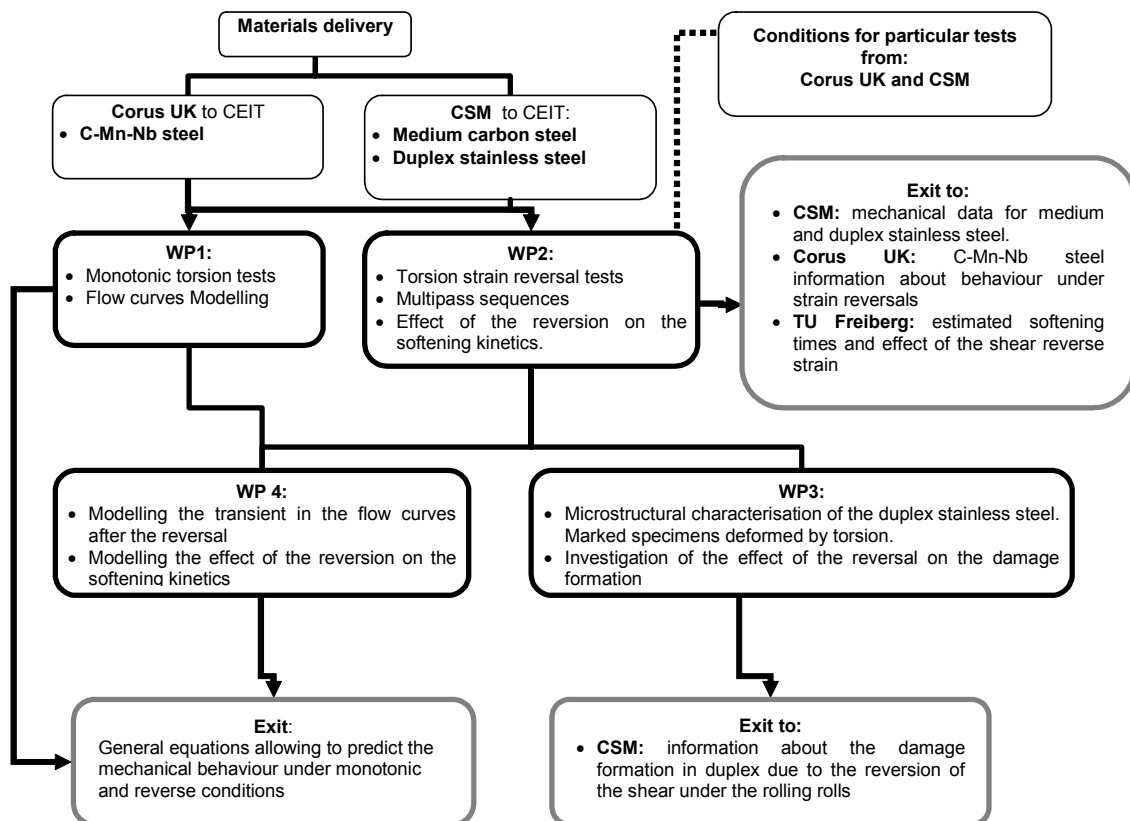
The main results that can be outlined from the present work are listed in the following:

- A model previously proposed by other authors has been applied successfully to model the hot torsion curves under monotonic conditions.
- The main parameters describing the transient on the flow curve after the reversion have been determined and related to the deformation conditions. The previous model for monotonic tests has been modified in order to describe the entire stress-strain curve (including the transient) when a reversion of the strain takes place during the deformation.
- The main characteristics of the effect of the transient after the reversal on the further softening kinetics have been determined. It has been shown that, depending on the magnitude of the reversion, the softening kinetics can be delayed or accelerated. The total absolute strain is not a good parameter to describe the effect of the reversion. The results have been discussed in terms of the stored energy.
- Stress and softening data have been produced under conditions specified by different partners, according to the simulations/modeling they were performing.

- Successive reversions lead to lower stress levels than those reached with the same total absolute strain applied under monotonic conditions.
- In duplex stainless steel, the stress-strain curve is fully reversible during the reversal. This cannot be related to the dissolution of the substructure, but to the internal stresses coming from the two phases. For small forward/reverse strains, the sliding at the interface is the main component and this one is fully reversible. The cracks become evident at the interface (mainly for the as-cast material) for intermediate strains.

A2.1 Links with other partners and workpackages.

The work carried out at CEIT has mainly used torsion facilities to perform multi pass deformation sequences involving or not reversions of the strain. The performed activities have been scheduled within the different workpackages in agreement with the work-programme agreed with the rest of the partners. The activities carried out under the different workpackages and the link with other partners is summarized in the following scheme.



The main results within the different workpackages are described in the following sections.

A2.2 WP1: Evaluation of constitutive models

A2.2.1 Introduction

The aim within this part of the work was to provide a model able to predict the hot working mechanical behaviour of a low carbon Nb microalloyed steel.

A2.2.1.2. Description of the model

It is well established that the mechanical behaviour under hot working conditions is strain rate ($\dot{\varepsilon}$) and temperature (T) dependent. Austenite has a relatively slow recovery rate, and dynamic recrystallisation activates for strains higher than a critical value, ε_c . The development of a model aimed to predict the shape of the stress-strain curve needs to consider separately the two strain ranges, $\varepsilon < \varepsilon_c$ and $\varepsilon \geq \varepsilon_c$.

- **Strain hardening and dynamic recovery.**

The model proposed by Laasraoui and Jonas [A2.1] has been applied to predict the flow curves. It is based on the following formulation:

$$\frac{d\rho}{d\gamma} = \left. \frac{d\rho}{d\gamma} \right|_{\text{stored}} - \left. \frac{d\rho}{d\gamma} \right|_{\text{recovery}} \quad \dots (A2.1)$$

with ρ , the dislocation density and γ the shear strain. The substitution of appropriate functions describing the stored energy and recovery terms and the integration of the resulting equation [A2.1] leads to the following expression for the flow stress, σ , as a function of the strain:

$$\sigma = \left[\sigma_s^{*2} + (\sigma_0^2 - \sigma_s^{*2}) e^{-\Omega \varepsilon} \right]^{0.5} \quad \text{for } \varepsilon < \varepsilon_c \quad \dots (A2.2)$$

The involved parameters are the yield stress σ_0 , the saturation stress attributable to the balance between strain hardening and dynamic recovery alone, σ_s^* , and a term, Ω , expressing the contribution of dynamic recovery. This term can be expressed as a function of the austenite grain size, d_0 , the strain rate and the temperature:

$$\Omega = A_{\Omega} d_0^{n_{\Omega}} \dot{\varepsilon}^{m_{\Omega}} \exp(Q_{\Omega}/RT) \quad \dots (A2.3)$$

with Q_{Ω} the activation energy. The saturation stress, σ_s^* can be expressed, as a function of the strain rate and the deformation temperature through the Zener-Hollomon equation:

$$\dot{\varepsilon} = A \left[\sinh(\alpha \sigma_s^*) \right]^n \exp(-Q/RT) \quad \dots (A2.4)$$

- **Dynamic recrystallisation range**

The model for the dynamic recrystallisation part of the curve leads to the following expression:

$$\sigma = \left[\sigma_s^{*2} + (\sigma_0^2 - \sigma_s^{*2}) e^{-\Omega \varepsilon} \right]^{0.5} - \left[\sigma_s^* - \sigma_{ss}^{drxn} \right] \left[1 - \exp \left(-k \left(\frac{\varepsilon - \varepsilon_p}{\varepsilon_{0.5} - \varepsilon_p} \right)^{n_D} \right) \right] \quad \dots \text{(A2.5)}$$

where σ_0 and σ_s^* are determined as before, n_D ranges between 1.3 and 1.5, $\varepsilon_{0.5}$ the strain for 50% softening by dynamic recrystallisation, σ_{ss}^{drxn} is the steady state stress and ε_p the peak strain:

A2.2.3 Experimental: monotonic torsion tests

Monotonic torsion tests have been performed on a C-Mn-Nb steel provided by Corus UK Limited with the composition shown in Table A2. 1. The deformation temperatures ranged in the interval 850 to 1200°C and the strain rates between 0.01 s⁻¹ and 1 s⁻¹. Prior to the deformation, the specimens were reheated at 1250°C, held there over 15 min and subsequently cooled down to the deformation temperature. The stress-strain curves were calculated using the method proposed by Fields and Backoffen [A2.2].

A2.2.4 Comparison with experiment

The above mentioned formulation has been applied to the experimental curves obtained from the monotonic torsion tests leading to the following expressions:

$$\dot{\varepsilon} = 1.51 \times 10^{11} \left[\sinh(0.012 \sigma_s^*) \right]^5 \exp(-325000 / RT) \quad \dots \text{(A2.6)}$$

$$\dot{\varepsilon} = 6.00 \times 10^9 \left[\sinh(0.068 \sigma_0) \right]^5 \exp(-325000 / RT) \quad \dots \text{(A2.7)}$$

$$\Omega = 8.7 \dot{\varepsilon}^{-0.08} e^{-9000 / RT} \quad \dots \text{(A2.8)}$$

It is to be noted that the values obtained for m (-0.08) and for Q_Ω (-9000 J/mol) are not far from those obtained previously by other authors [A2.1,A2.3]. Additionally:

$$\sigma_{ss}^{drxn} = 83.34 \sinh^{-1} \left[\left(Z / 3.4 \cdot 10^{11} \right)^{0.22} \right] \quad \dots \text{(A2.9)}$$

$$\varepsilon_p = 2.7 \cdot 10^{-3} Z^{0.19} \quad \dots \text{(A2.10)}$$

This model predicts reasonably well the experimental stress strain curves for $\varepsilon < \varepsilon_c$, as can be seen in the examples shown in Fig. A2.1.

A2.3 WP2: Experimental investigation of constitutive behaviour

The main objectives of this part of the work are listed in the following:

- To investigate, the effect of the reversion on the stress and the recrystallisation kinetics, using a hot torsion allowing the application of relatively large strains without the instabilities problems usually associated to tension and compression.
- To simulate the effect of the shear strain reversion taking place in the roll gap.

- To investigate the effect of the shear strain reversion on the damage formation during hot working on a duplex stainless steel.
- To provide experimental results to other partners from specially designed torsion tests.

A2.3.1 Experimental: Selection of the materials

The steels used to perform the torsion strain reversal tests are listed in Table A2. 1.

• **C-Mn-Nb steel**

The same C-Mn Nb steel provided by Corus UK Limited has also been used in order to investigate the effect the reversal has on the flow curves and on the recrystallisation kinetics during the transient after a reversion of the strain. The composition of this steel was selected as close as possible to the one used in a previous project [A2.4].

• **Medium carbon steel**

A medium carbon steel provided by CSM has been used to obtain stress-strain curves under specific conditions defined by CSM in order to generate useful data to be used into the FEM modelling. The aim of performing these tests was to get complementary information that could be applied for modelling mechanical behaviour under complex deformation mode. The chemical composition of this steel is given by CSM in Table A3.1 of Appendix 3.

• **Duplex stainless steel**

A duplex stainless steel provided by CSM has been used to perform torsion tests involving different strain paths. This steel belongs to the cast identified as 426877 with a composition shown in Table A2. 1. As-cast and wrought materials have been used to perform the tests. The wrought material corresponds to the one identified as MAT 1 that was industrially hot rolled (ILVA) to a 77% reduction. Previous works carried out at CEIT [A2.5-A2.11] have shown that the as cast and the wrought materials present different microstructures and behave differently when hot deformed. It was aimed in the present work to investigate how these materials behave under strain reversal tests.

A2.3.2 Experimental: Strain reversal tests

The same torsion machine used previously allows to perform multipass torsion tests, including the reversion of the strain between two consecutive passes. The experimental work procedure applied for each material is described in the following.

• **C-Mn-Nb steel**

Strain reversal torsion tests have been carried out with this steel in order to:

- Obtain additional information on the effect the reversion has on the mechanical response of the steel under a broad range of conditions.
- Investigate the effect the reversion has on the recrystallisation kinetics.

To reach the first of these objectives, two consecutive passes were applied with a reversion of the twist between both. The flow curves were recorded and compared. In order to

investigate the effect of the reversion on the recrystallisation, the following sequences were applied:

Two pass tests. Two forward passes, separated by a given holding time were applied, according to the schedule shown in Table A2.2. The second pass was used to determine the fractional softening caused by the holding time at the deformation temperature.

Three pass tests. The first pass was applied in the forward and the second and the third ones in the reverse. A holding time was applied between the second and the last pass. The deformation conditions and amount of reversal applied in the second pass are shown in Table A2.2.

The static fractional softening was determined from the stress-strain curves by the 2% offset method:

$$R_{0.02} = \frac{\sigma_{pre} - \sigma_{0.02}^*}{\sigma_{pre} - \sigma_{0.02}} \quad \dots (A2.11)$$

where $\sigma_{0.02}$ and $\sigma_{0.02}^*$ are the 2% offset yield stress in the fully recrystallised steel and after the holding time respectively and σ_{pre} is the pre-stress before unloading for holding. This method provides results that correlate well with the recrystallized fraction obtained by quantitative metallography for microalloyed steels [A2.12].

In all the cases, the material was preheated before deformation to 1250°C, held there during 15 min and cooled to the deformation temperature.

- **Medium carbon steel**

The torsion tests performed on this steel can be classified as follows:

- Monotonic tests and reversal tests: carried out in order to obtain constitutive experimental data to feed the FEM programs (pure shear).
- Multipass tests to determine the softening taking place during given holding times applied after different strain paths.
- Multipass reversal tests: These tests include a sequence of different low strain passes with a series of strain reversals. They are aimed to simulate the sequence of reversions of the shear component that takes place under the rolls. The sequence simulates what happens during the rolling passes, assuming no softening takes place from one pass to the next one.

The tests were carried out according to the conditions predefined by CSM, Table A2.3. The samples were reheated at a temperature of 1050°C and held there for 2 min, cooled to the deformation temperature and held there for 15 s before the start of the deformation sequence.

Multipass torsion sequences involving combinations of pre-strain, reverse strain and interpass holding times were applied, according to the details shown in Table A2.3. The deformation pass applied after the holding period allowed determining the softening fraction according to the 2% offset method. The sequence, see Table A2.4, was scheduled in order to simulate the reversion of the shear that takes place in the roll gap (no holding time) and

between the passes (holding time after the reversion). Short interpass holding times (0.3-0.9 s) were applied that have been estimated to be comparable to those in the rolling mill. The sequence in Table A2.5 was repeated four times for every set of multipass conditions and two times for those conditions denoted by * in Table A2.4.

- **Duplex stainless steel**

The material was heated to the deformation temperature and held there for 1 h. The conditions for these tests are shown in Table A2.6. Two strain rates (0.1 and 1 s^{-1}), two deformation temperatures (950 and 1100°C) and different forward and reverse strains were applied.

All along the present work, the absolute strain and absolute stress have been used, for simplicity, to plot the stress-strain curves.

A2.3.3 Results on the C-Mn-Nb steel

- **Effect of the reversal on the stress-strain curves**

The graphs in Fig. A2.2 show the stress strain curves obtained for the reversals carried out after 0.3 and 0.7 pre-strains at 1100 and at 850°C . It can be seen that just after the reversal, the stress increases rapidly to reach a level slightly lower than the pre-stress. This is followed by some stagnation of the stress and defines a plateau that is not so clearly distinguished than on the tests carried out previously at a lower strain rate (0.1 s^{-1}) [4].

For the case of the test carried out at 1100°C with a pre-strain within the dynamic recrystallisation range ($+0.7$), the curve after the reversal reaches directly the stress corresponding to the steady state, without passing through the peak. This indicates that some interaction between the dynamic recrystallisation and the reversion takes place when the pre-strain in the forward excess ε_c . All these results are in agreement with the behaviour previously observed for another steel of similar composition [4,13].

- **Effect of the strain path on the static softening.**

The fractional softening during the holding time after different reversed strains has been determined according to the method explained in Section A2.3.2. In the absence of softening, it is expected that the stress after holding would increase to reach the pre-stress level. In the case of full recrystallisation, the steel would behave independently of the previous strain path. The curves after a partial softening show an intermediate behaviour.

Pre-strain $> \varepsilon_c$

At 1100°C for a pre-strain in the forward of $\varepsilon_f=0.7$, which is within the dynamic recrystallisation range, followed by different amounts of reversal, the reversal is interfering with dynamic recrystallisation, as commented before. In this case, the recrystallisation kinetics after the holding time is independent on the reversed strain, probably due to the activation of metadynamic recrystallisation. However, it has to be mentioned that the softening times after the reversals are slightly longer than those corresponding to the monotonic pre-strain of 0.7.

Pre-strain $< \varepsilon_c$

This is the range that has particularly been investigated in the present work. The experimental results are summarised in Fig. A2.3, for different deformation conditions and strain paths. The following rules are fulfilled:

- The full reversion of the strain (+0.3; -0.3), (+0.15; -0.15) produces shorter recrystallisation times than those reached after applying only the pre-strain (+0.3) (+0.15), but longer than expected for a monotonic strain equal to the total absolute strain.
- The same is true for some intermediate partial reversion of the strain for example (+0.3, -0.2) for which the softening curve is between those corresponding to the strain path (+0.3;-0.3) and (+0.3) curves.
- For small reverse strains, the recrystallisation time is approximately the same than that for the pre-strain. For example, the softening curves for the test (+0.3) and (+0.3;-0.1) approximately overlap.
- For even lower pre-strains, (+0.15;-0.06) the recrystallisation times are longer than those for a monotonic pre-strain of 0.15.

This is in agreement with previous results [A2.14-A2.16]. The longest softening times are obtained for reverse strains close to the end of the plateau.

- **Additional tests**

Some additional tests were performed under conditions defined by Corus UK in order to investigate the softening during a short time (5 s) and different strain paths. It was aimed to approach the strains and times involved in industry and in laboratory trials. The obtained results are shown in Table A2.7 and Table A2.8. The time required to get 50% softening and the n value for the Avrami equation are also indicated for certain conditions.

A2.3.4 Results on the medium carbon steel

The medium carbon steel used to perform the torsion tests was the same supplied by CSM also to the University of Oulu for the tension-compression tests. The aim of performing these tests was to get complementary information that could be applied by CSM for the modelling of the mechanical behaviour under the roll gap.

- **Monotonic tests**

The obtained flow curves for the monotonic are shown in Fig. A2.4. The main parameters defining the curves: strain to the peak, ε_p , peak stress, σ_p , the strain for the start of the steady state, ε_{ss} , and the steady state flow stress, σ_{ss} , are shown in Table A2.9.

- **Strain reversal tests and interpass softening**

The obtained stress-strain curves for different strain paths and holding times are shown in Fig. A2.5 for a deformation temperature of 1050°C. The monotonic curves and those corresponding to different strain paths or holding times are superposed on the same graphs. Excepting for the monotonic curves, between the first and the second pass a reversion was applied (no holding time) a holding time took place between the second and the last pass (no reversion). For the deformation sequence, (+0.1, -0.1), none of the applied holding times: 1 s (curve in black) or 10 s (curve in grey) produces a significant softening and the last pass curve approximately superimpose.

When higher forward and reverse strains are applied, some softening is observed, as can be observed in the second graph of the same figure. This last has been obtained for the deformation sequence +0.2, -0.2, leading to 24% and 66% softening after 1 s and 10 s respectively. All the obtained results with this steel are summarised in Table A2.10.

- **Multipass strain reversal tests**

The obtained flow curves with the multipass sequences, carried out at a deformation temperature of 950°C and strain rates of 0.1 s⁻¹ and 1 s⁻¹ are shown in Fig. A2.6. For comparison, monotonic flow curves have also been plotted in the same figure. A sequence of consecutive reversed passes, leads to stress in the passes significantly lower than that for a monotonic test to the same total equivalent strain.

When comparing the flow curve envelope for the multipass test, σ_{sf} , to the monotonic flow curve, σ_{mono} , see Fig. A2.7, it is possible to define a variable (σ_{gap}) for the same total absolute strain as:

$$\sigma_{gap}(\varepsilon) = \sigma_{mono}(\varepsilon) - \sigma_{sf}(\varepsilon) \quad \dots (A2.12)$$

The value of σ_{gap} depends on the pre-strain and on the deformation conditions. High temperatures lead to low values, but high strain rates raise the σ_{gap} .

- **Main results and conclusions**

- The sequences involving forward and reverse strains of 0.1 need holding times between passes higher than 10 seconds to initiate static softening. For strains of 0.2, a significant amount of softening is reached for these same holding times.
- For small cyclic strains, the stress level in successive passes remains significantly lower than that for a monotonic test to the same total equivalent strain.

A2.3.5 Results on the duplex stainless steel

From previous works carried out at CEIT [A2.5-A2.11], the following characteristics can be outlined that differentiate as cast and wrought duplex stainless steel materials:

As- cast material

- Microstructure formed by large primary δ -ferrite grains delimited by ridges of austenite. Inside the δ -ferrite grains Widmanstätten austenite plates are present. An example of the microstructure is shown in Fig. A2.8.
- During the solidification δ -ferrite forms from the liquid and the austenite precipitates in solid state during further cooling. As a result, most of the ferrite-austenite interfaces are partially coherent due to the K-S orientation relationship between both phases.
- During deformation, the austenite blocks oriented almost at random within the as-cast microstructure, rotate to align according to the imposed deformation conditions. The main deformation mechanisms are: preferential shearing of the ferrite softest phase, interface sliding and austenite block rotation.
- Coherence has an important role in hot ductility given that interface sliding is not possible on coherent interfaces and has been observed to concentrate at the incoherent portions, leading to the formation of cracks at these locations.
- The torsion curve has the aspect shown in Fig. A2.8. Three regions are observed: elastic region, followed by strain hardening stage to reach a peak, followed by a monotonic decrease of the stress.
- The softening is attributed to the formation of damage and no signs of dynamic recrystallisation in the microstructure of the austenite are observed for any of the performed tests.

Wrought material

- Microstructure formed by alternate austenite and ferrite parallel bands, Fig. A2.8.
- The coherency at the interface between ferrite and austenite has been lost, as a result of the lattice rotations taking place during deformation.
- The main deformation mechanism is preferential shearing of the ferrite softest phase and interface sliding.
- The sliding at the interface distributes more homogeneously among the different interfaces and damage formation is reduced significantly. The torsion curves present the shape shown in Fig. A2.8: The stress rises to reach a peak, followed by a slight decrease of the stress. This last cannot be attributed to dynamic recrystallisation, since no signs of this last is observed in the microstructure. At the peak, the ferrite has developed subgrains, while the austenite remains nearly undeformed, which is an indication of the strain partitioning taking place between the two phases.
- **Behaviour of the as-cast material under a reversion of the strain:**

The Fig. A2.9(a) shows the stress-strain curve corresponding to a test involving a reversion of the strain after a pre-strain of about 0.06, which is within the range of those taking place within the roll gap during rolling. The test has been carried out at a strain rate of 0.1 s^{-1} . The

curve after the reversal does not exhibit the plateau commonly observed in single-phase materials. Instead, some parabolic strain hardening is observed to reach the same stress level than before the reversion. The Bauschinger effect manifests as a lower yielding after the reversal than the pre-stress level just before the reversion. A similar behaviour is also observed for higher pre-strains before the reversal, Fig. A2.9(b).

At higher strain rates, no significant changes are observed, due to the reversion of the strain both for small forward and reverse strains, as can be seen for the curves in Fig. A2.10, corresponding to a test carried out following a deformation sequence involving cyclic forward and reverse straining. The same type of behaviour is observed, independently of the reversion and the place of the curve within the cycle. However, for larger pre-strains (>0.2), Fig. A2.9(c), some differences can be outlined: reverse stress after the reversion exhibits a small peak and the stress level is slightly higher than that before the reversion. At pre-strains >0.5 , Fig. A2.9(d), some softening is observed after the reversion, probably associated to some damage formation, before fracture.

- **Behaviour of the wrought material under a reversion of the strain:**

In Fig. A2.11(a), the flow curves corresponding to two tests carried out at two different temperatures and the same strain rate of 0.1 s^{-1} are shown. For the two cases, the reversion of the strain has been applied for a pre-strain of about 0.06. For these small deformations, the obtained curves are not very different from those reached with the as-cast microstructure, Fig. A2.8.

In Fig. A2.11(a-c), the strain reversal curves for higher pre-strains in the forward are shown. It can be seen that for a pre-strain of 0.2, a transient showing a yield stress plateau is observed. After the transient, following the reversion, the strain hardening activates again and, as it happens for the case of single-phase materials, the shape of the curve, after the appropriate shift along the strain axis, becomes similar to that for monotonic tests. For a strain reversal after a higher pre-strain, the plateau does not manifest after the reversal, and there is a complete reversibility of the shape of the curve. The graph in Fig. A2.11(d) shows the flow curves for strain reversal tests carried out at 1000 and 1050°C and a strain rate of 1 s^{-1} . The shape of the curves is very similar to the one obtained for the as-cast material, Fig. A2.9 the same happens for the cyclic tests, Fig. A2.10.

A2.4 WP3: Material characterisation

The work carried out within this work package has mainly concentrated on the investigation of the effect of the reversal on the duplex stainless steels in connection with the damage formation. Some observations have also been done to investigate the interaction between the reversal and the dynamic recrystallisation in the C-Mn-Nb steel.

A2.4.1 Results of the microstructural characterization. Duplex stainless steel

The main characteristics of the as-cast and wrought duplex stainless steel microstructures have been described in the previous section. The main goal of this part of the work has been to analyse the behaviour of the duplex microstructure under a reversion of the strain. For this, some torsion tests have been performed on specimens previously polished and marked with some scratches.

- **Effect of the reversion on the wrought duplex microstructure**

The original alternate elongated ferrite and austenite bands parallel to the torsion axis that are characteristic of the wrought microstructure remain almost unchanged after a small forward and reverse strain, as can be seen in Fig. A2.12. The scratches remain nearly continuous and no damage formation is observed for the strain reversal tests carried out for a pre-strain of 0.06, followed by the full reversion of the strain. The resulting microstructure at the subsurface of the torsion specimen exhibits nearly imperceptible signs of deformation.

The micrographs in Fig. A2.13 show the marked surface of specimens following multipass cyclic reverse torsion tests, for two different strain rates. It can be seen that, after 6 cycles, the scratches maintain the continuity across the ferrite-austenite interface and that there is no signs of damage formation, as was previously observed when a higher degree of reversion was applied [A2.4,A2.17]. The micrograph in Fig. A2.14 shows the duplex microstructure at the subsurface of the torsion specimen without no significant evidence of a substructure development.

- **Effect of the reversion on the as-cast duplex microstructure:**

When small forward-reverse strains are applied to the as-cast material, the scratch remains straight Fig. A2.15. However, large strains lead to the formation of damage. As an example, in Fig. A2.16, the torsion specimen surface is shown for a test deformed in the forward to a strain of 0.56 and in the reverse to fracture, (flow curve in Fig. A2.9). Some displacement of the scratches is clearly evident associated to some sliding at certain δ/γ interfaces, probably the incoherent ones that are expected to slide more easily than K-S orientation related ones.

- **Reversion and damage formation**

For the wrought material, the reversibility of the flow curve after the reversal is clearly evident for low and high pre-strains, Fig. A2.17. The austenite blocks act as hard “fibres” into the soft ferrite matrix and rotate to accommodate to the imposed torsion. On reversing the strain, the “fibres” return to the original position, before starting the rotation in the reverse. This rotation is the main responsible for the linear strain hardening in the flow curves of the wrought material, Fig. A2.8. The difference, $\Delta\sigma$, between the stress just after the reversal and the minimum stress during the following transient can be associated to the work required to rotate the “fibres”. This contribution is relatively small for low pre-strains but becomes important for a large deformation, see Fig. A2.18. But for all the conditions, the possible contribution of the dissolution/rebuilding of a dislocation substructure during the transient is masked by the internal stresses coming from the two phases with significantly different mechanical properties.

The consecutive forward and reverse strains around 0.06 are not able to enhance damage formation, at least under pure shear. However, higher strains lead to the early decohesion of the ferrite-austenite interface in wrought steels. For the as-cast material, the sliding is nearly impossible on the most part of the ferrite-austenite, and concentrates on the incoherent interfaces. This is true for monotonic and also for reverse strains.

A2.4.2 Results of the microstructural characterization. C-Mn-Nb steel:

It has been observed that when the strain reversal takes place within the dynamic recrystallisation range of strains, the behaviour of the flow curve after the reversion is clearly modified. In order to investigate the interaction between DREX and SREV, different tests

have been carried at 1050°C and strain rate 0.1 s⁻¹. Two monotonic tests up to $\varepsilon=0.38$ (peak strain) and $\varepsilon=0.76$ (halfway between the peak and the onset of the steady state), followed by water quenching. Tests involving different reverse strains after the peak strain has been reached have been performed: $\varepsilon=0.38/-0.38$, $\varepsilon=0.38/-0.76$. The dynamic recrystallisation seems to progress during the reversal that explains the low strain hardening during the second part of the transient observed when the pre-strain is within the dynamic recrystallisation range [A2.13].

A2.5 WP4. Formulation of constitutive laws

After the reversal of the strain for this same steel, a deviation of the flow curve from the behaviour in a monotonic test is observed (Bauschinger effect).

A2.5.1 Modelling the transient after a reversion of the strain for the C-Mn-Nb steel

The mechanical behaviour after the reversal can be analysed by means of the experimental assessment of several parameters. The first one is the relative drop in stress just after changing the direction of twist, σ_{plat} . The second one refers to the length of the plateau, $\varepsilon_{\text{plat}}$ (defined as the strain range where $\delta\sigma/\delta\varepsilon \approx 0$). Finally, the third selected parameter, $\varepsilon_{\text{BAUS}}$, or Bauschinger strain is the strain required to reach the same stress value than that at pre-strain, σ_f .

For the tests carried out under constant absolute Z, it has been found that the relative drop in stress is approximately independent of the test conditions, excepting the pre-strain, leading to the following relations between σ_{plat} and the pre-stress, σ_{pre} , see Fig. A2.19(a):

$$\sigma_r = 0.95\sigma_f \quad \dots \text{ (A2.13)}$$

The $\varepsilon_{\text{plat}}$ is mainly dependent on the prestrain, ε_{pre} , Fig. A2.19(b):

$$\varepsilon_{\text{plat}} = 0.33\varepsilon_{\text{pre}}^{3/4} \quad \dots \text{ (A2.14)}$$

The Bauschinger strain ($\varepsilon_{\text{BAUS}}$) appears to follow the same trend, leading to:

$$\varepsilon_{\text{BAUS}} = 0.62\varepsilon_{\text{pre}}^{3/4} \quad \dots \text{ (A2.15)}$$

For the modelling of the flow curve after the reversal, $\sigma_r(\varepsilon)$, the following equation has been used:

$$\sigma_r(\varepsilon) = \left[\sigma_s^{*2} + (\sigma_0^2 - \sigma_s^{*2}) e^{-\Omega(\varepsilon - \varepsilon_{\text{Baus}})} \right]^{0.5} + B\varepsilon^{-b} \quad \dots \text{ (A2.16)}$$

with σ_0 and σ_s^* determined through Equations (A2.5) and (A2.6), $b=4.41$ and B being a function of Z and the pre-strain. Two examples of the predictions obtained with the model are shown in Fig. A2.20. Such an equation is very sensitive to the value of B and some of the above Equations (A2.13 to A2.15) need to be used as filters, when using the mentioned formulation in order to avoid incoherent results.

A2.5.2. Strain rate change simultaneous to the reversal

The parameters used to characterise the transient, σ_{plat} , $\varepsilon_{\text{plat}}$ and $\varepsilon_{\text{BAUS}}$ appear as being independent on the deformation conditions. Additional tests in which the strain rate is modified in sign and absolute value at the same time have been carried out.

Drop in strain rate + reversal: Several tests involving a drop in the strain rate of one order of magnitude ($0.1 \text{ s}^{-1} \rightarrow 0.01 \text{ s}^{-1}$) have been performed. The decrease in the strain rate makes the Bauschinger effect more pronounced and the drop in stress is higher than that at constant absolute strain rate. Additionally, the length of the plateau becomes larger. This length increases as the pre-strain does, but at a faster rate than that stated for constant absolute strain rate tests. These results are not surprising since, at lower strain rate, the excess of mobile dislocation density produced by the decrease in strain rate superimpose to those due to the dilution of the substructure by the reversal.

Rise in strain rate + reversal: Torsion strain reversal tests at 950°C involving an order of magnitude increases in strain rate ($0.01 \text{ s}^{-1} \rightarrow 0.1 \text{ s}^{-1}$) at the same time the strain is reversed have been performed. The main results obtained from these tests are the near complete disappearance of any plateau and the change of the sign of the variation of stress after the reversion/rate change. Additionally, the transient seems to shorten as the prestrain gets larger. The balance between the need for more mobile dislocation density to sustain the rise in the strain rate and the dissolution due to the reversal could explain the described observation.

A2.5.3 Modelling the relation between the softening kinetics and the strain path

For the monotonic tests the softening times decrease as the pre-strain increases. By reversing the strain, the recrystallisation times depend on the reversed strain, Fig. A2.21. All the softening curves obtained in this work can satisfactorily be fitted by a Johnson-Mehl-Avrami-Kolmogorov equation of the form:

$$X = 1 - \exp \left[-0.693 \left(\frac{t}{t_{0.5}} \right)^n \right] \quad \dots \text{(A2.17)}$$

with X , the fraction softened, t , the holding time, $t_{0.5}$, the time required for 50% softening and $n=1$. For monotonic conditions, $t_{0.5}$ can be estimated, as a function of the strain, the strain rate and the temperature, using the equations available in the literature for Nb microalloyed steels. However, $t_{0.5}$ depends on the strain path and it is difficult to find a parameter taking into account such effect. It is clear that the total absolute strain is not the appropriate parameter, Fig. A2.21. Lindh et al [A2.18], proposed that the effective strain can be expressed in terms of the permanent and redundant strain in the way:

$$\varepsilon_{\text{eff}} = \varepsilon_{\text{per}} + 0.65\varepsilon_{\text{red}} \quad \dots \text{(A2.18)}$$

This expression is applicable for large reverse strains, but it is unable to predict the observed delay on the softening kinetics during the transient.

The driving pressure for static recrystallisation after hot working, F_R° , can be assumed to decrease with time due to the early consumption during recrystallization of the regions with the highest stored energy, the variations in the mobility of the migrating front due to the segregation of the solutes and/or to recovery processes in the non recrystallised areas. A

quite general formula [A2.19-A2.21] relating the decay with time of the driving pressure for recrystallization can be written in the form:

$$F_R(t) = F_R^0 \left(1 + \frac{t}{\tau_1}\right)^{-u} \quad \dots (A2.19)$$

with τ_1 a relaxation time and $0 < u < 1$. Assuming site saturation, the recrystallised fraction can be expressed as [A2.22]:

$$X_v = 1 - \exp \left[-N_{\text{Rex}} \left(\int_0^t m(t) F_R(t) dt \right)^3 \right] \quad \dots (A2.20)$$

Assuming the driving pressure for nucleation and for growth are approximately the same during the early stages of recrystallization and preferential nucleation at grain boundaries $N_{\text{Rex}} = \lambda S_v (F_R^0)^2$, being λ a geometric factor and S_v the austenite specific grain boundary area before recrystallization starts. Assuming the temperature dependent mobility, m , remains constant with time, this equation can be rewritten as:

$$X_v = 1 - \exp \left[-\lambda S_v m^3 (F_R^0)^5 \left(\int_0^t \left(1 + \frac{t}{\tau_1}\right)^{-u} dt \right)^3 \right] \quad \dots (A2.21)$$

To make, after integration, the time dependence in this Equation consistent with the one found experimentally ($n = 1$) when applying Equation (A2.17) to the present steel: $u = 1 - \frac{n}{3} = \frac{2}{3}$. For $\tau_1 \ll t$, the integration leads to:

$$0.693 \frac{t}{t_{0.5}} = \lambda S_v m^3 (F_R^0)^5 \left(\frac{\tau_1}{1-u} \right)^3 \left(\frac{t}{\tau_1} \right)^{3(1-u)} \quad \dots (A2.22)$$

leading to:

$$F_R^0 = \left[\frac{0.693}{27 * \lambda m^3 \tau_1^2} \frac{1}{S_v t_{0.5}} \right]^{1/5} \quad \dots (A2.23)$$

In this equation, F_R^0 and S_v depend on the strain and strain path; F_R^0 The variation of S_v has been determined [A2.23], assuming the full reversibility of the shape of the grains during the reversal [A2.24]. The following expression:

$$\frac{F_{R(\varepsilon)}^0}{F_{R(\varepsilon_f)}^0} = \left[\frac{S_{v(\varepsilon_f)}}{S_{v(\varepsilon)}} \frac{t_{0.5(\varepsilon_f)}}{t_{0.5(\varepsilon)}} \right]^{1/5} \quad \dots (A2.24)$$

allows making an estimation of the relative driving pressure for recrystallization for different strain paths. The estimation is refereed for each set of experimental results to a particular forward strain ε_f , Fig. A2.22. The plain symbols correspond to the tests involving a reversion of the strain. It can be seen that, in general, during the transient after the reversal a decrease in the estimated relative driving pressure is observed.

The driving pressure for recrystallisation is defined by the stored energy that can be expressed as the sum of two terms the first one takes into account the dislocation density at the cell interiors and the second one the misorientation across the cell boundaries. The reversion of the strain is expected to modify the balance between the two contributions. The stress relates to the total dislocation density, but not to their arrangement. However, if one plots the pre-stress before holding, as a function of the $(S_{vt0.5})^{-1/5}$, Fig. A2.23, a linear relation is obtained for each temperature and strain rate that is applicable to both monotonic and strain reversal tests. The use of this type of representation only gives approximate softening times following a reversion but it is not able to predict the maximum in the recrystallisation times observed at the end of the plateau.

A2.6 Conclusions

• **Strain reversal tests**

- The tests carried out in the two carbon steels confirm, for a broad range of conditions, the mechanical behaviour previously deduced in a previous ECSC project [A2.4]. The reversion of the strain produces a transient on the flow curves that can be associated to the partial dissolution of the substructure and the rebuilding of a new one compatible with the new deformation conditions.
- The transient is characterised by a plateau (no strain hardening) the main parameters defining it: stress and length (strain) depend on the pre-strain and initial and final strain rates.

• **Modelling. C-Mn-Nb steel**

- The model proposed in Reference (A2.1) can be successfully applied to predict the monotonic flow curves of the Nb microalloyed steel in the present work. This model can be modified by using an additional term to describe the transient after the reversal for constant absolute strain rate and temperature. In such a case, the transient stress and strain only depend on the pre-stress or the applied pre-strain.
- The effect of the reversal on the softening kinetics has clearly been outlined in the present work. The partial dilution of the substructure during the reversion leads to a decrease of the stored energy and a delay in the softening. The longest softening times are observed at the end of the plateau (highest dilution). The accurate modelling of such effect needs the use of physically based models that are not available at this moment. The instantaneous pre-stress before the holding can be used to make a rough estimation of the delay produced by the reversion on the softening kinetics, but this parameter has not enough sensibility to describe with accuracy such an effect.

• **Duplex stainless steel**

- The differences between the as-cast and wrought materials in terms of the answer to the reversal are not significant at low strains, but increase with increasing the pre-strain, as a result of the reversibility of the flow curve.

- In duplex stainless steel, the most of the cases, there is a lack of the stress stagnation after the reversion commonly observed for single-phase materials. For the wrought material, the reversibility of the flow curve after the reversal is clearly evident for low and high pre-strains. The austenite blocks act as hard “fibres” into the soft ferrite matrix and need to rotate to accommodate to the imposed torsion. On reversing the strain, the “fibres” return to the original position, before starting the rotation in the reverse. This rotation is the main responsible for the linear strain hardening in the flow curves of the wrought material and is the responsible of difference, $\Delta\sigma$, between the stress just after the reversal and the minimum stress during the following transient. This contribution is relatively small (for low pre-strains but becomes important for a large deformation).
- In a duplex microstructure, the contribution of the dissolution/rebuilding of a dislocation substructure during the transient is masked by much strong contributions coming from the presence of the two phases with significantly different mechanical properties.
- The reversal involving forward and reverse strains around 0.06 is not able to enhance damage formation. However, higher strains were observed to lead to the early decohesion of the ferrite-austenite interface in wrought steels. For the as-cast material, the sliding is nearly impossible on the most part of the ferrite-austenite, as a result of the semi-coherent character they have. The interface sliding, which is an important source of damage in duplex steels, mainly concentrates on the incoherent interfaces. This is true for monotonic and also for reverse strains.
- **Cyclic forward/reverse torsion tests**
- The stress level reached in subsequent passes involving the reversion of the strain is significantly lower than that corresponding to a monotonic test to the same total absolute strain.

A2.7 References

- A2.1 Laasraoui, A and Jonas, J J: Metallurgical Transactions A, 1991, Vol. 22A, 1545-1558.
- A2.2 Fields, Jr D S and Backoffen, W A: Proc. ASTM, vol. 57, 1957, 1259.
- A2.3 Yoshie, A, Morikawa, H and Onoe, Y: Trans. ISIJ, 1987, Vol. 27, 425.
- A2.4 Zhou, M, Morris, P F, Husain, Z, Wiklund, O, Terziyski, T, Peura, P, Karjalainen, L P, Bianchi, J H, Gutierrez, I, Bartolomé, R, Piñol-Juez, A and Iza-Mendia, A: The Effect of Strain Reversal and Strain-Time Path on Constitutive Relationships for Metal Rolling/Forming Processes. Technical Steel Research, EUR 19891 EN, Office for the Official Publications of the European Communities, Luxembourg, 2001.
- A2.5 Piñol, A, Iza, A, Gutierrez, I and Urcola, J J: CMMC 96, Key Engineering Materials, Vol. 127-131, Ed. M. Fuentes et al., Trans Tech. Publications, Switzerland, 1997, pp1025-1032.

- A2.6 Iza, A, Piñol, A, Gutierrez, I and Urcola, J J: CMMC 96, Key Engineering Materials, Vol. 127-131, Ed. M. Fuentes et al., Trans Tech. Publications, Switzerland, 1997, 1033-1040.
- A2.7 Iza-Mendia, A, Piñol-Juez, A, Urcola, J J and Gutierrez, I: Metall. Mater. Trans., 1998, Vol. 29A, 2975.
- A2.8 Piñol-Juez, A, Iza-Mendia, A and Gutierrez, I: Metall. Mater. Trans.. A, 2000, Vol. 31A, 1671.
- A2.9 Piñol-Juez, A, Iza-Mendia, A and Gutierrez, I: 13th European Conf. on Fracture (ECF 13) Proc. Eds. Fuentes, M, Elices, M, Martin-Meizoso, A and Martinez-Esnaola, J M, Elsevier, 2000, Paper Ref. 11U.293.
- A2.10 Piñol-Juez, A, Iza-Mendia, A and Gutierrez, I: International Conf. on "Thermomechanical processing: Mechanics, Microstructure and Control, 23-26 June 2002, the Univ. of Sheffield, England.
- A2.11 Piñol-Juez, A, Iza-Mendia, A and Gutiérrez, I: In European Research Conf. on "Plasticity of Materials. Fundamentals, Modelling and Application" European Science Foundation, 25-30, April, Granada, 1998.
- A2.12 Fernández, A I, López, B and Rodríguez-Ibabe, J M: Scripta Mater., 1999; Vol. 40: 543.
- A2.13 Bartolomé, R, Astiazarán, J I, Iza-Mendia, A and Gutierrez, I: Proc. Thermomechanical Processing of Steels, IOM Communications, May 2000, London, 221.
- A2.14 Bartolomé, R, Jorge-Badiola, D, Astiazarán, J I and Gutiérrez, I: Mat. Sci. and Eng. A, 2003, Vol. 344, 340.
- A2.15 Jorge-Badiola, D, Bartolomé, R, Martín, S and Gutiérrez, I: International Conf. on "Thermomechanical processing: Mechanics, Microstructure and Control, 23-26 June 2002, the Univ. of Sheffield, England.
- A2.16 Jorge-Badiola, D and Gutiérrez, I: Acta Mater., 2004, Vol. 52, 333.
- A2.17 Zhou, M, Gutierrez, I, Bianchi, J H, Wiklund, O, Schmidtchen, M and Karjalainen, P: Constitutive Modelling Complex Loading in Metal Forming Processes, ECSC Project 7210.PR/291. 2002 annual report.
- A2.18 Lindh, E, Hutchinson, B and Veyama, S: Scripta Metal and Mater, 1993, Vol. 29, 347.
- A2.19 Furu, T, Marthinsen, K and Nes, E: Mat. Sci. Tech., 1990; 6: 1093.
- A2.20 Humphreys, F J and Hatherly, M: "Recrystallization and Related annealing Phenomena", Pergamon, Oxford, 1995. p127.
- A2.21 Sellars, C M: 3rd Int. Conf. on "Recrystallization and Related Phenomena", Ed. McNelley, T R, Monterey Institute of Advanced Studies, Monterey, USA, 1997. p81.

- A2.22 Zurob, H S, Hutchinson, C R, Brechet, Y and Purdy, G: *Acta mater.*, 2002; Vol. 50, 3075.
- A2.23 Cotrina, E, Iza-Mendia, A, López, B and Gutiérrez, I: *Metall Mater. Trans. A*, 35, 2004, 93.
- A2.24 Farag, M M, Sellars, C M and McG. Tegart, W J: "Deformation under hot working conditions" (Special Rep. N° 108), Iron Steel Institute, London, 1968. p60.

Table A2. 1: Composition (wt. %) of the steels used into the project

	C	Si	Mn	P	S	Al	Nb	V	N	Mo	Cu	Cr	Ni
C-Mn-Nb	.15	.30	1.42	.012	.002	.037	.033	.011	.007	.003	.012	0.02	0.03

	Cr	Ni	Mo	Mn	Si	P	S	C	N
Duplex	23.1	4.83	0.22	1.30	0.51	0.024	0.001	0.025	0.095

Table A2.2: Test conditions for the multipass sequences applied to the C-Mn-Nb steel

Temperature (°C)	Strain rate (s ⁻¹)	Strain path	Interpass time (s) 1st-2nd pass/2nd-3rd pass
850	1	Monotonic	
		0.3	0/from 5 to 500
		0.7	0/from 3-100
		0.3/(-0.3,-0.4)/-1	0/5
		0.7/(-0.1,-0.7)/-1	0/5
875	0.1	0.3/-1	0
		0.55/-1	0
900	0.1	Monotonic	
		0.6/-0.6	0
930	0.1	0.4/-1	0
		0.5/-1	0
		0.65/-1.2	0
950	0.1	0.1/-1	0
		0.2/-1	0
	0.01	0.1/-1	0
		0.15/-1	0
1050	0.1	Monotonic	
		0.4/from 0 to -0.4	0
	0.01	From 0.05 to 0.16/-0.9	0
1100	0.1	0.12/1	From 6 to 300
		0.15/1	From 25-400
		0.2/1	From 16 to 300
		0.25/1	From 8 to 200
		0.2/-0.06/-1	0/ from 25 to 400
		0.2/-0.1/-1	0/ from 35 to 1000
		0.2-0.15/-1	0/ from 8 to 200
		0.2/-0.2/-1	0/ from 8 to 3000
	1	Monotonic	
		0.3/1	From 0.4 to 40
		0.7/1	From 0.4 to 5
		0.3/(-0.1,-0.3)/-1	0/ from 0.4 to 5
		0.7/(-0.1,0.7)/-1	0/ 5
1150	0.1	0.1/1	0/ from 25 to 200
		0.15/1	0/from 8 to 100
		0.2/1	0/ from 3 to 200
		0.15/-0.06/-1	0/ from 3 to 1000
		0.15/-0.1/-1	0/ from 3 to 1000
		0.15/-0.15/-1	0/ from 3 to 1000
1200	0.01	From 0.03 to 0.12/ -0.8	0

Table A2. 3: Torsion test conditions for the medium carbon steel

Deformation temperature (°C)	Strain rate (s ⁻¹)	First pass	First holding time (s)	Second pass	Hoding time(s)	Third pass
950	0.1	2	-	-	-	-
950	1	2	-	-	-	-
1000	1	3	-	-	-	-
1050	0.1	0.8	-	-	-	-
1050	1	3	-	-	-	-
1050	10	10	-	-	-	-
1050	1	+0.1	0	-0.1	1	-1
1050	1	+0.1	0	-0.1	10	-1
1050	1	+0.1	0	-0.9	-	-
1050	1	+0.2	0	-0.2	1	-1
1050	1	+0.2	0	-0.2	10	-1

Table A2. 4: Test conditions for the multipass sequences. medium carbon steel

Temperature (°C)	Strain rate (s ⁻¹)	Strain path
950	0.1	Monotonic
		Multipass strain-reversal*
	1	Monotonic
		Multipass strain-reversal
1000	0.1	Multipass strain-reversal*
		Monotonic
	1	Multipass strain-reversal
		Monotonic
1050	0.01	Multipass strain-reversal
	0.1	Monotonic
		Multipass strain-revesal
	1	Monotonic
		Multipass strain-reversal

Table A2.5: Forward (+), reverse (-) strains and interpass holding times applied for the multipass sequences in the previous table

ε_1	$t_{p1}(s)$	ε_2	$t_{p2}(s)$	ε_3	$t_{p3}(s)$	ε_4	$t_{p4}(s)$
+0.1	0	-0.1	0.3	+0.1	0	-0.1	0.9

Table A2.6: Conditions of the tests carried out with the duplex stainless steel

Material	Deformation temperature (°C)	Strain rate (s ⁻¹)	Pre-strain	Reverse strain	Sequences of cyclic reversions
As -cast	1000	0.1	0.06	0.06	
	950	0.1	0.2	to fracture	
	1000	1	0.06	0.06	6 passes
	1000	1	0.56	to fracture	
	950	1	0.2	to fracture	
Wrought	1050	0.1	0.06	0.06	
	1000	0.1	0.06	0.06	
	950	0.1	0.22	to fracture	
	950	0.1	0.8	to fracture	
	1100	1*	0.17	0.17	10 passes
	1050	1	0.06	0.06	
	1000	1*	0.06	0.06	
	1000	1	0.06	0.06	6 passes
	1000	1	0.12	0.12	
	1000	1*	0.14	to fracture	
	1000	1	0.17	0.17	10 passes
	1000	1	0.17	to fracture	
	1000	1	0.19	0.19	
	1000	1	0.8	to fracture	
	1000	1	0.87	0.87	
	900	1*	0.17	0.17	10 passes

The results of the tests marked with * were reported in the final report of a previous project [A2.8].

Table A2. 7: $t_{0.5}$, n , t_p and fractional softening after 5 s. holding for different thermal cycles and monotonic prestrains: $\dot{\varepsilon} = 1 \text{ s}^{-1}$.

Preheating temperature (°C)	Deformation temperature (°C)	Prestrain	$t_{0.5}$ (s)	n	t_p (s)	Fractional softening
1300	1100	+0.3	2.1	1	-	0.73
1200	1100		2.0	1	-	0.81
1300	850		78.7	0.8	98	0.24
1300	1100	+0.7	0.6	1	-	0.96
1200	1100		0.5	1	-	-
1300	850		10.2	0.7	33	0.35

Table A2.8: $t_{0.5}$, n , t_p and fractional softening after 5 s. Holding for different thermal cycles and monotonic prestrains: $\dot{\varepsilon} = 1 \text{ s}^{-1}$

Preheating temperature (°C)	Deformation temperature (°C)	Strain path	$t_{0.5}$ (s)	n	Fractional softening
1300	1100	+0.3/-0.1	2.3	1	0.77
		+0.3/-0.2	1.5	1	0.90
		+0.3/-0.3	1.1	1	0.93
		+0.7/(-0.7,-0.1)	1.5*	-	0.89
	850	+0.3/(-0.4,-0.3)	-	-	0.21/0.18
		+0.7/-0.1	16*	-	0.24
		+0.7/-0.25	7*	-	0.40
		+0.7/-0.7	2*	-	0.85

* estimated

Table A2. 9: Main parameters obtained from the monotonic flow curves.
Medium carbon steel.

Deformation temperature (°C)	Strain rate (s^{-1})	ε_p	σ_p (MPa)	ε_{ss}	σ_{ss} (MPa)
950	0.1	0.4	108	1.0	97
950	1	0.7	152	1.6	133
1000	1	0.63	129		114
1050	0.1	0.3	67	0.7	61
1050	1	0.45	103	1.0	93
1050	10	1.3	160	2.3	145

Table A2. 10: Reversal test conditions and the obtained softening fraction.
Medium carbon steel.

Deformation temperature (°C)	Strain rate (s^{-1})	Prestrain	Reverse strain	Holding time (s)	Softening fraction (%)
1050	1	0.1	0.1	1	0
1050	1	0.1	0.1	10	8
1050	1	0.2	0.2	1	24
1050	1	0.2	0.2	10	66

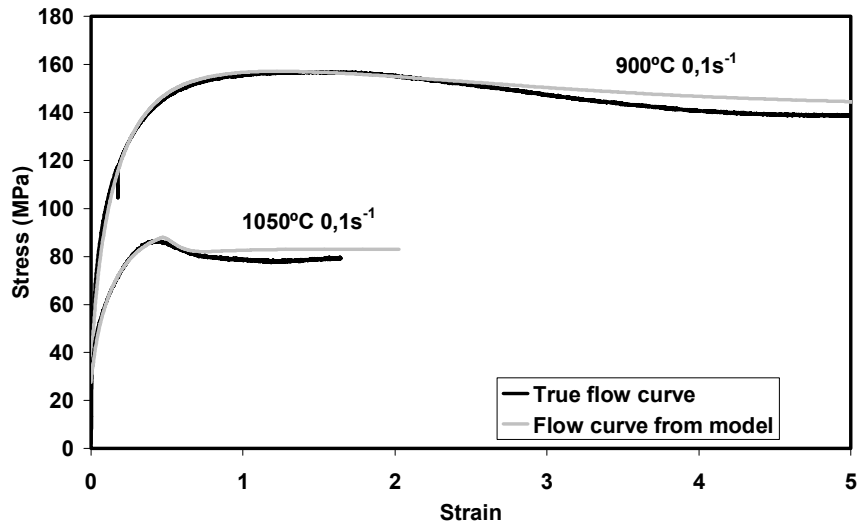
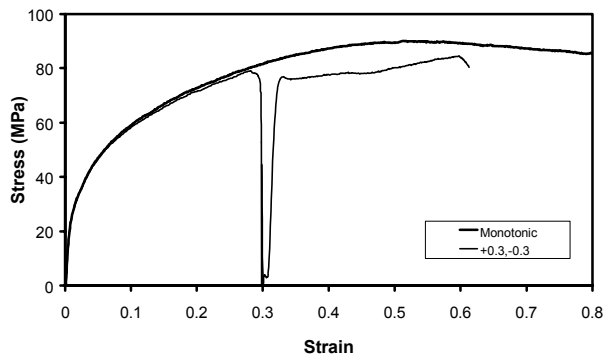
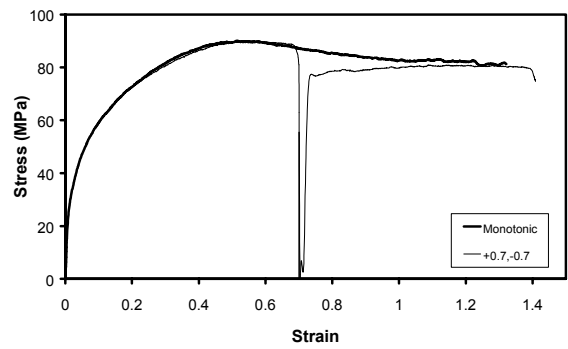


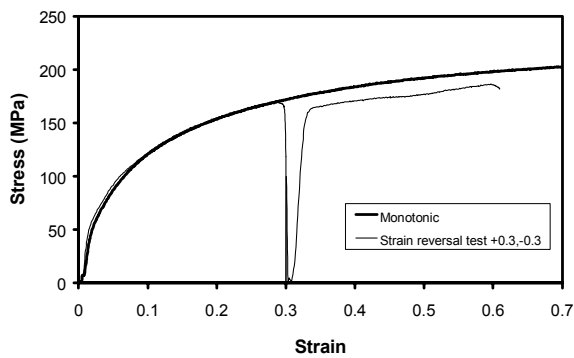
Fig. A2.1: Comparison between the experimental torsion curves and the predictions from the model



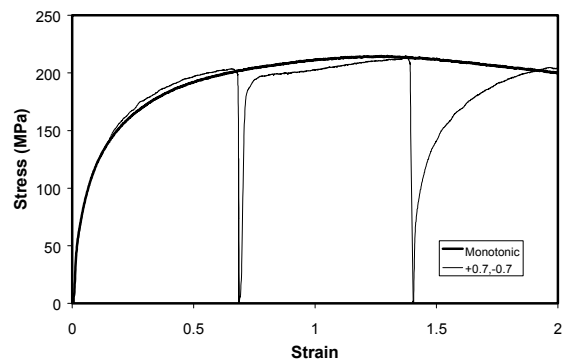
(a)



(b)



(c)



(d)

Fig. A2.2(a-d): Stress-strain curves obtained at with different strain paths. Strain rate: 1 s^{-1} and. (a) and (b) 1100°C and (c) and (d) 850°C

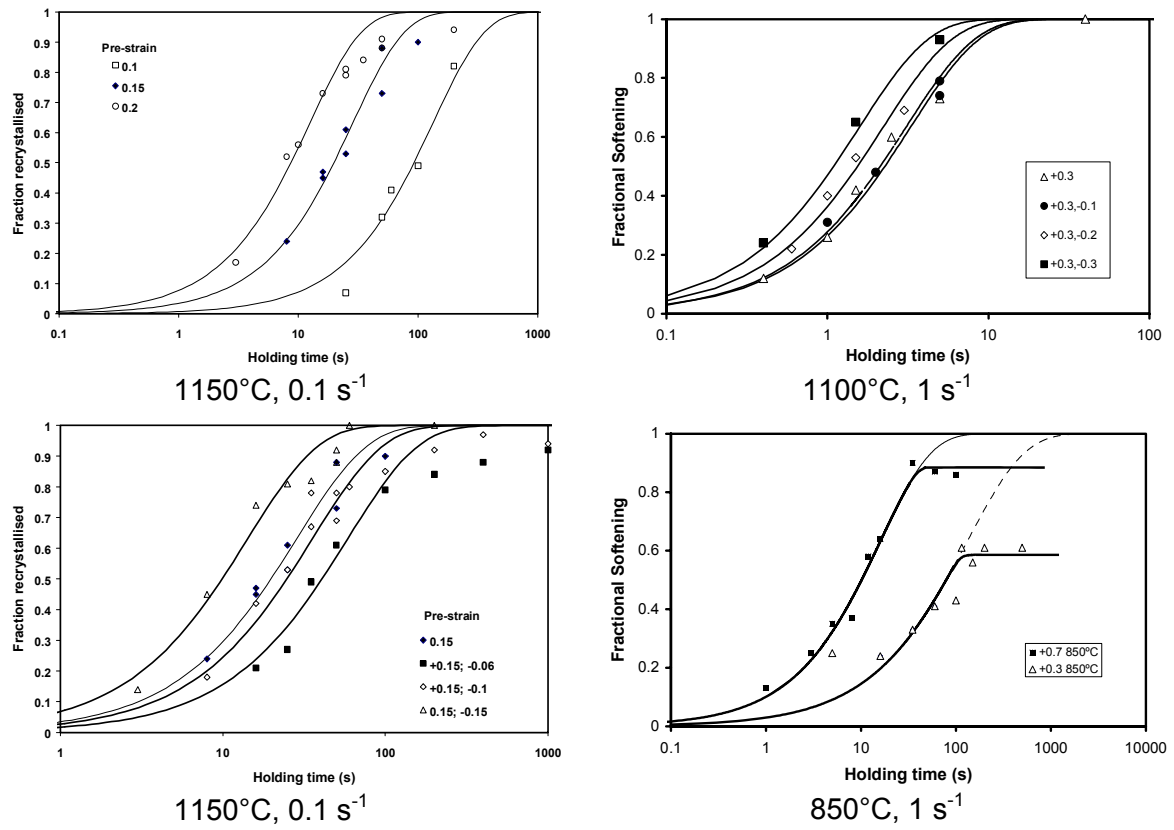


Fig. A2.3: Fractional softening for different strain paths. C-Mn-Nb steel.

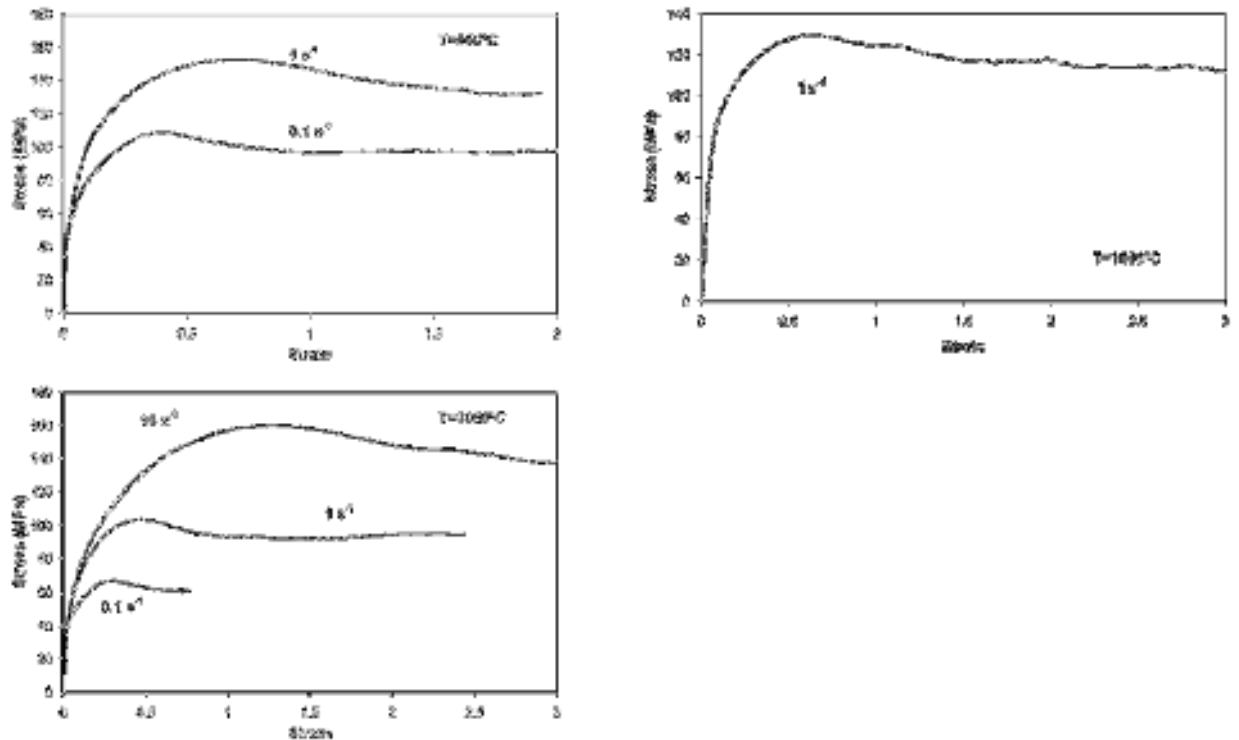
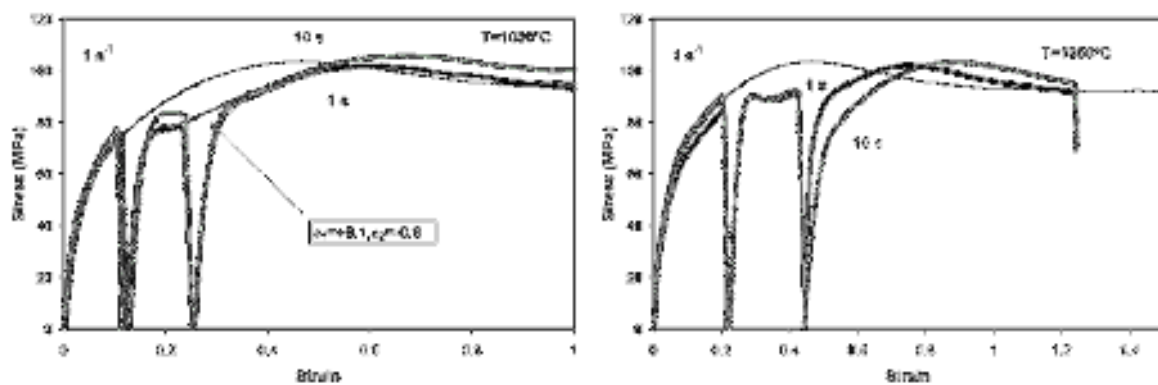


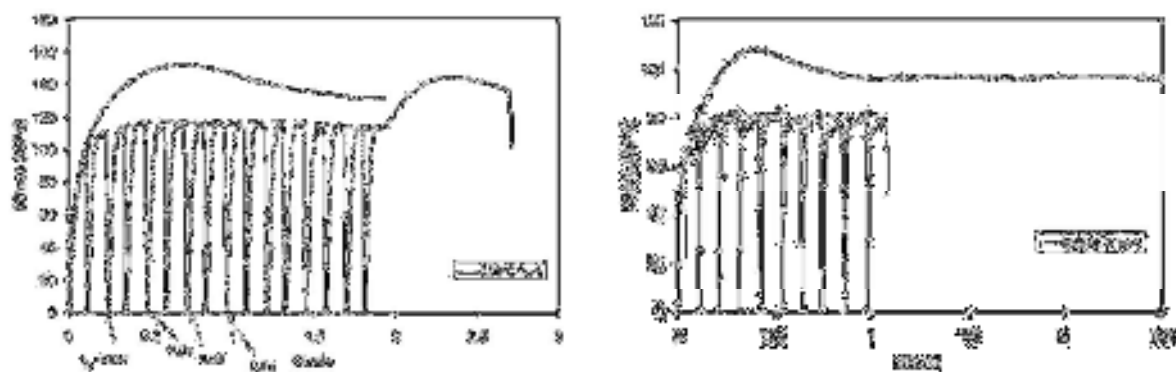
Fig. A2.4: Medium carbon steel monotonic flow curves



(a)

(b)

Fig. A2.5(a and b): Strain-stress curves for strain reversal tests. Deformation sequences: (a) $\varepsilon=+0.1, -0.1$ (b) $\varepsilon=+0.2, -0.2$. Medium carbon steel.



(a)

(b)

Fig. A2.6(a and b): multipass reversal curves and monotonic tests at 950°C and strain rate (a) 1 s^{-1} , (b) 0.1 s^{-1} .

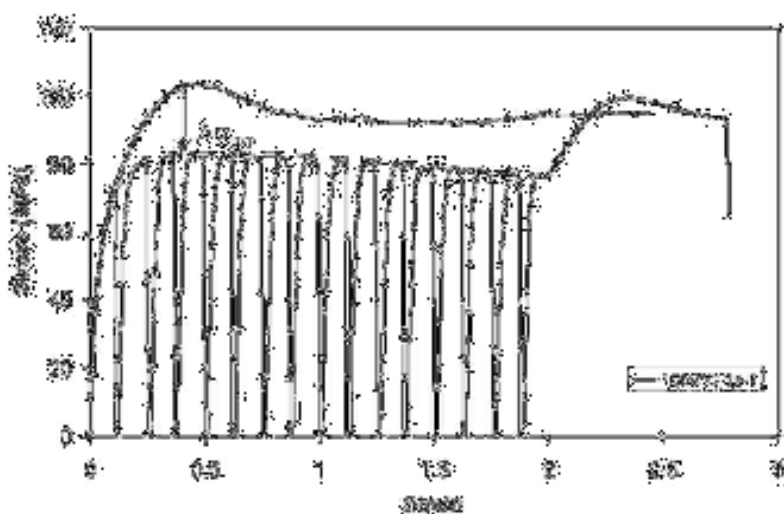


Fig. A2.7: Determination of σ_{gap} using the monotonic and multipass reversal tests

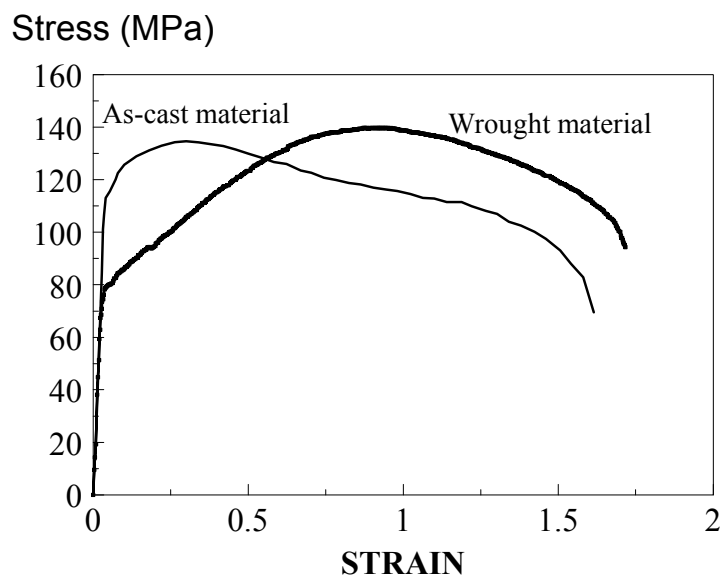
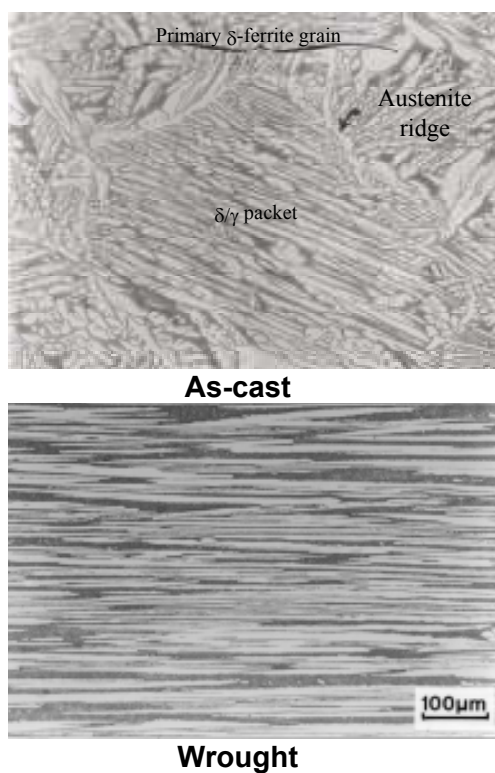
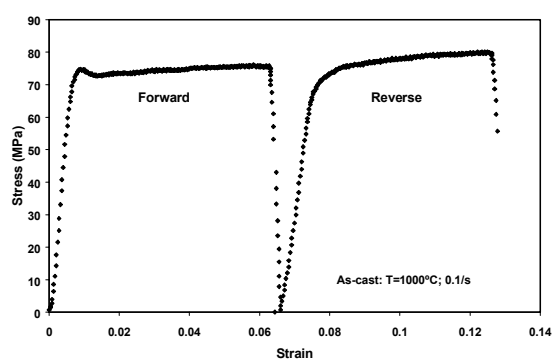
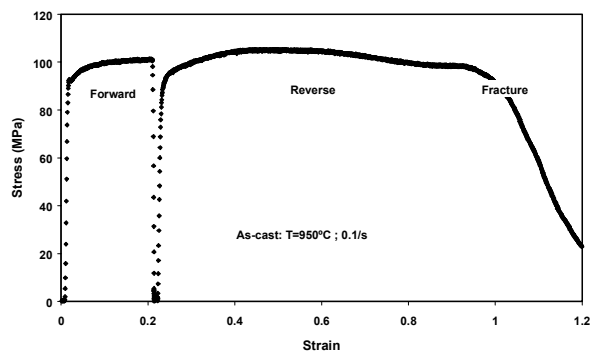


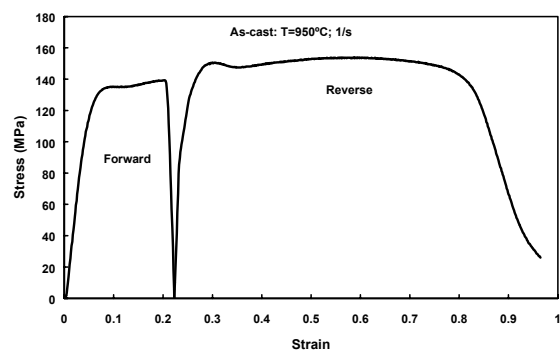
Fig. A2.8: As cast and wrought duplex stainless steel microstructures and torsion curves at 1000°C and 1 s⁻¹



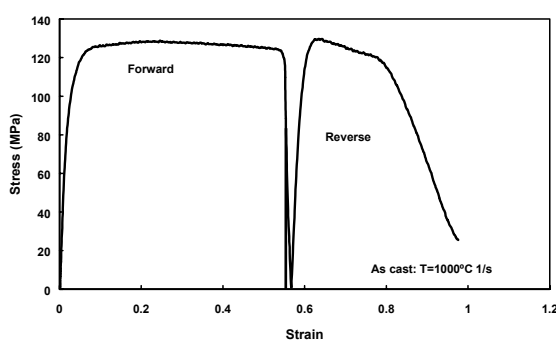
(a)



(b)

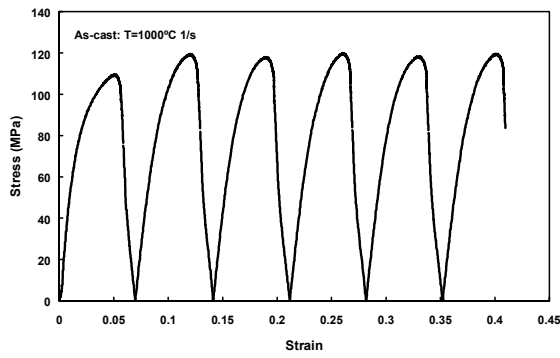


(c)

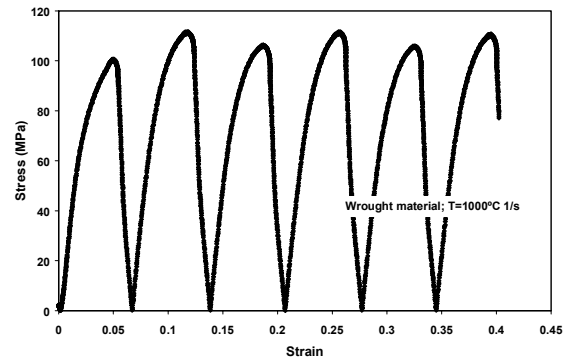


(d)

Fig. A2.9(a-d): As-cast duplex stainless steel: Strain reversal torsion curves for different strain rates and pre-strains.

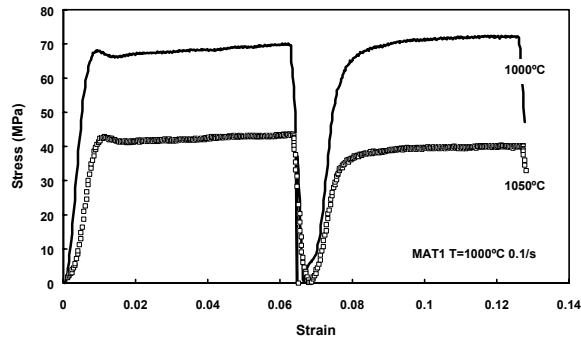


As-cast

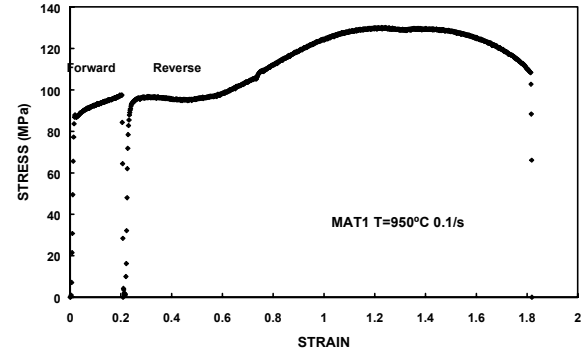


Wrought

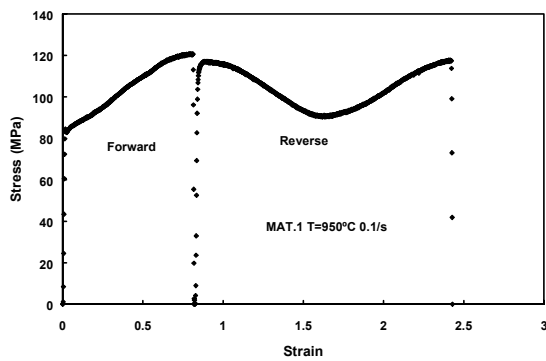
Fig. A2.10: As-cast duplex stainless steel: stress strain curve for a multipass torsion test including the cyclic reversion of the strain



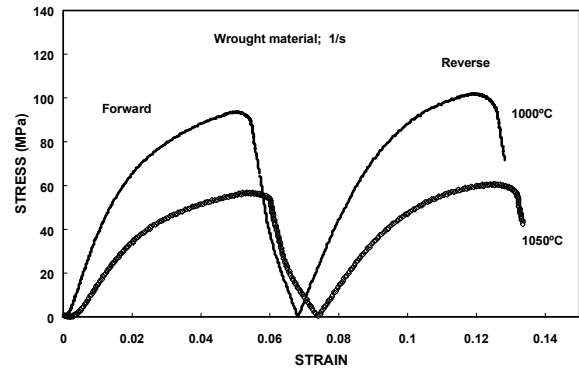
(a)



(b)

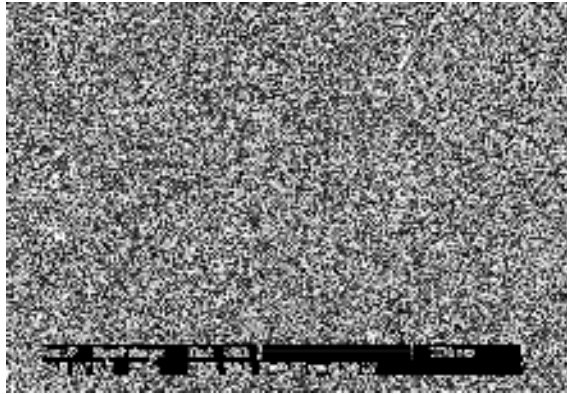


(c)

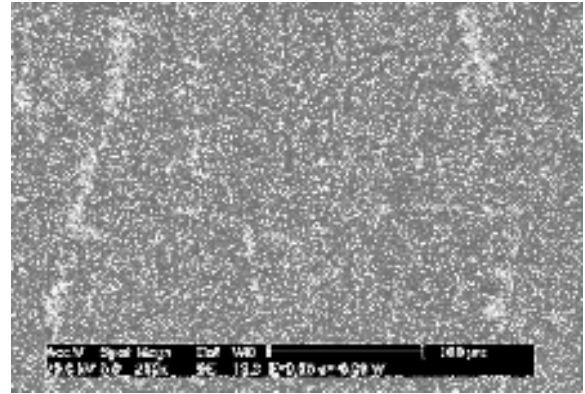


(d)

Fig. A2.11(a-d): Wrought duplex stainless steel. Strain reversal torsion curves for different deformation temperatures and pre-strains.

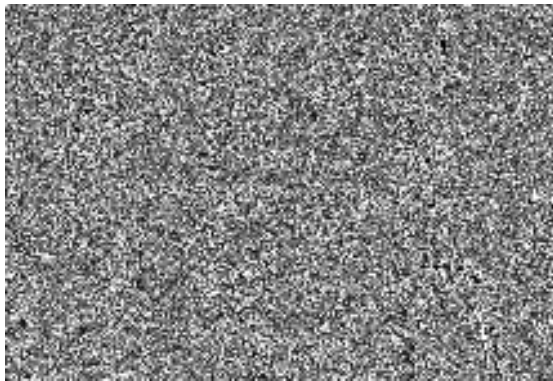


(a)

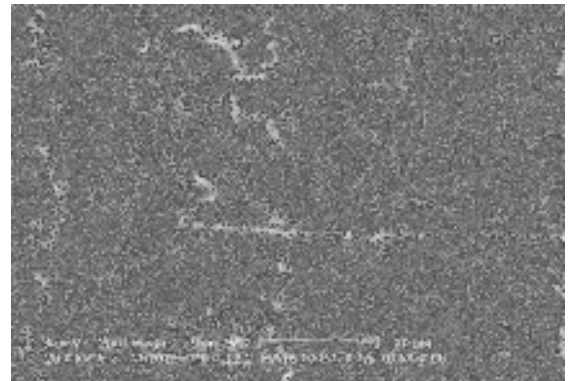


(b)

Fig. A2.12(a and b): marked torsion specimen surface. Torsion test at 1000°C and 1 s⁻¹. Pre-strain =0.06, reverse strain = 0.06. (a) directly after deformation and (b) after some additional polishing.



(a)



(b)

Fig. A2.13: marked torsion specimen surface. Cyclic 6 pass torsion tests carried out at 1000°C. Pre-strain =0.06, reverse strain = 0.06. (a) 1 s⁻¹ and (b) 0.1 s⁻¹.

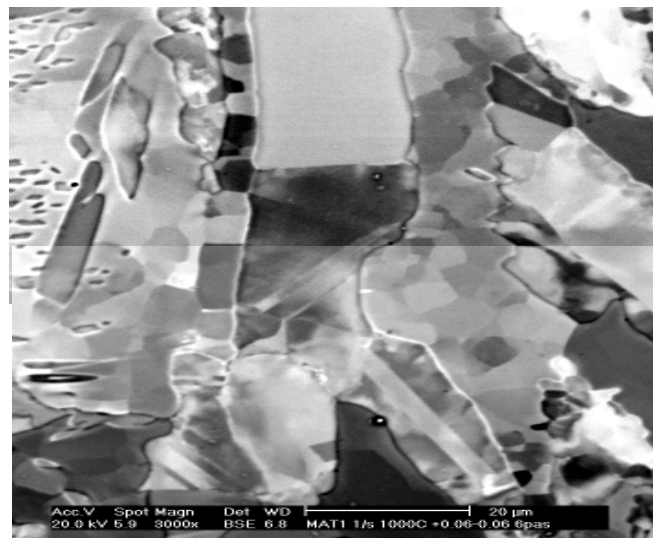


Fig. A2.14: SEM micrograph at the subsurface of the torsion specimen after a 6 pass cyclic test at 1000°C and 1 s⁻¹. Forward/reverse strain: 0.06.

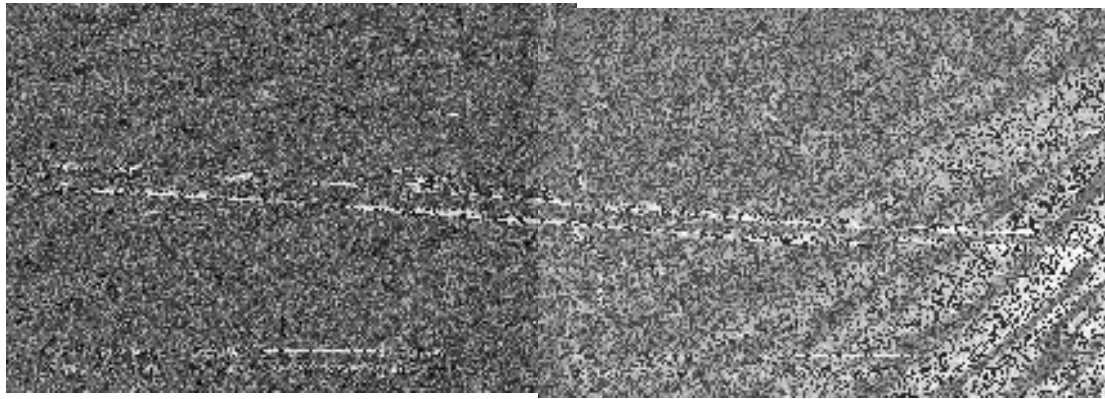


Fig. A2.15: Marked torsion specimen surface. Test at 1000°C and 1 s⁻¹. As-cast material. Pre-strain =0.06, reverse strain = 0.06.

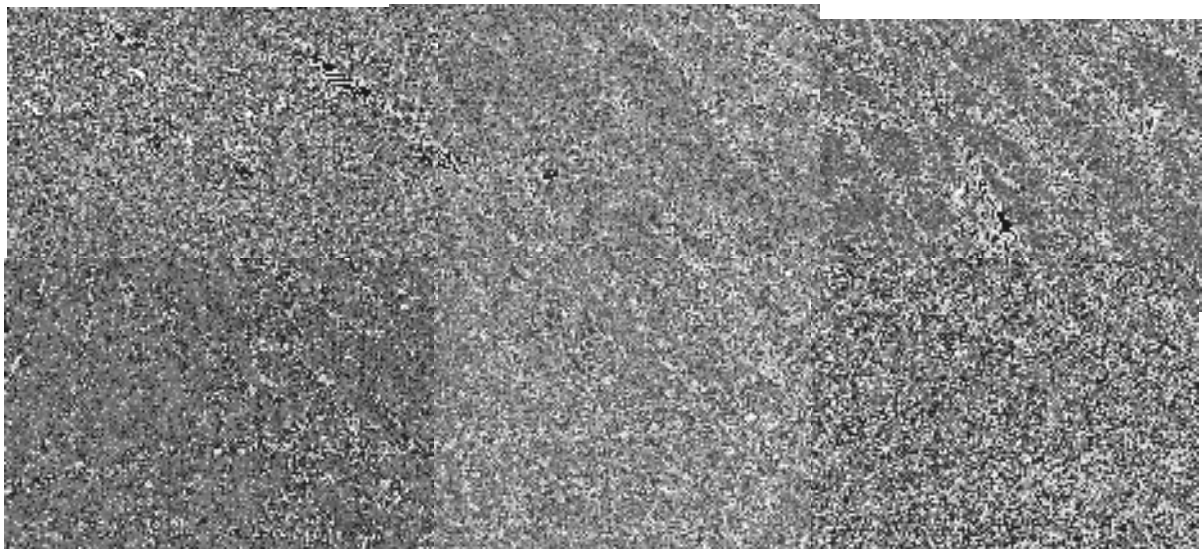


Fig. A2.16: Marked torsion specimen surface. Test at 1000°C and 1 s⁻¹. As-cast material. Pre-strain =0.56, reverse strain to fracture.

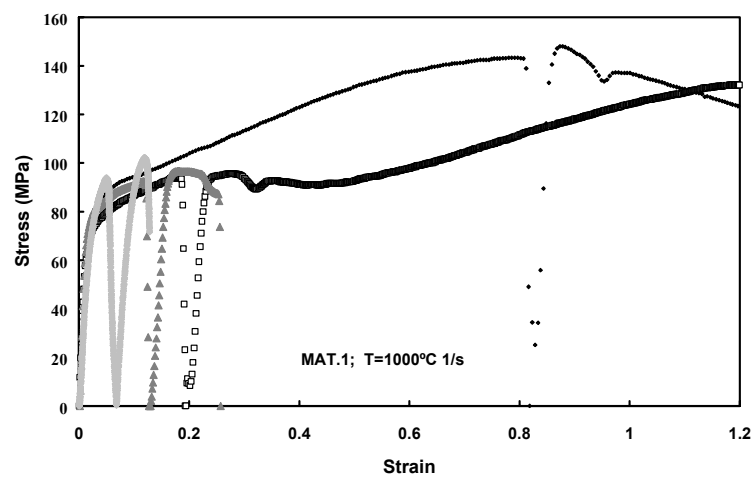


Fig. A2.17: Comparison between the flow curves obtained for different strain reversal tests

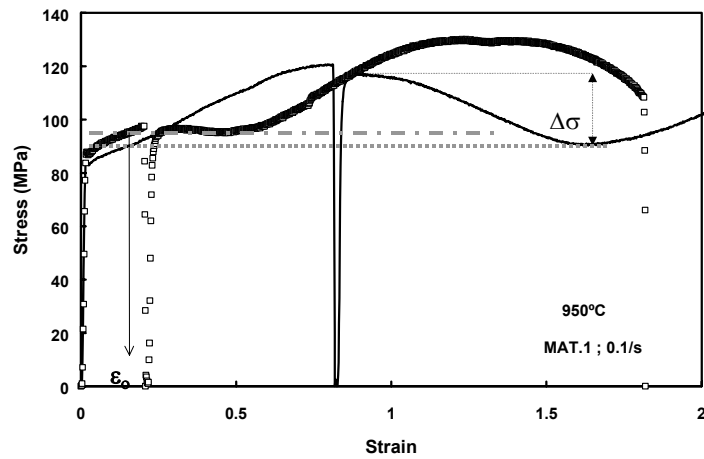
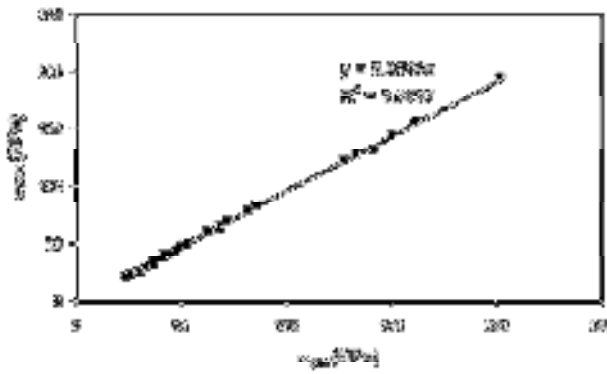
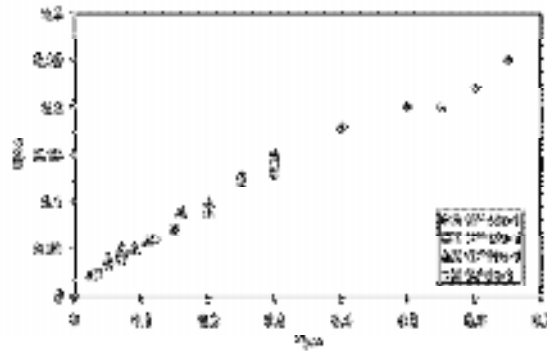


Fig. A2.18: Stress contribution associated to the rotation of the austenite "fibers" during torsion



(a)



(b)

Fig. A2.19: Transient parameters for different Z values

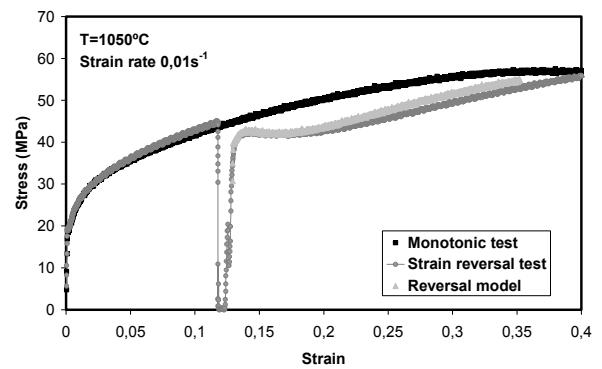
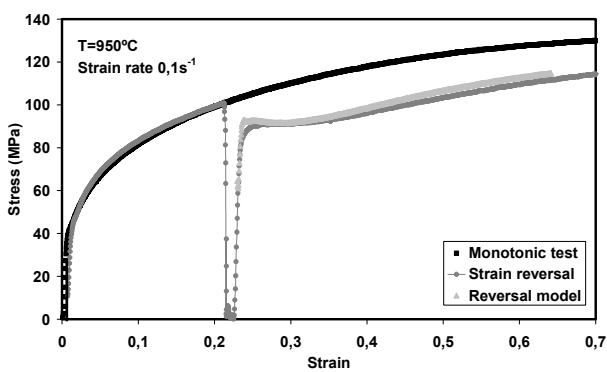


Fig. A2.20: Predicted curves after the reversion compared to experiment

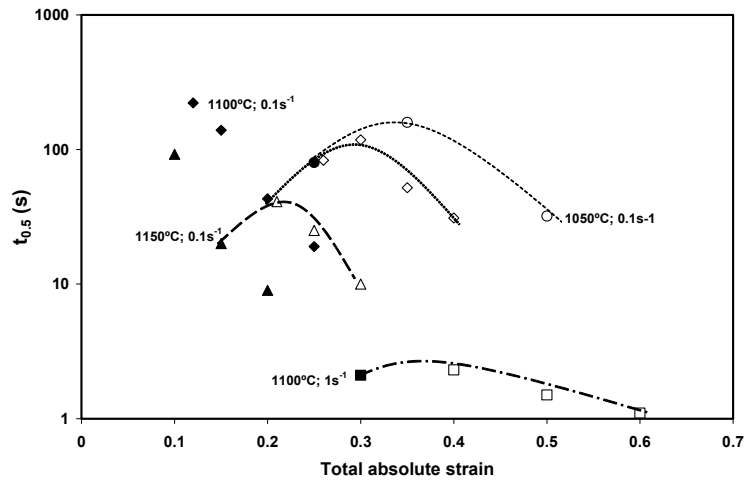


Fig. A2.21: Time to 50% softening, as a function of the total absolute strain. Symbols: black for monotonic, plain for strain reversal.

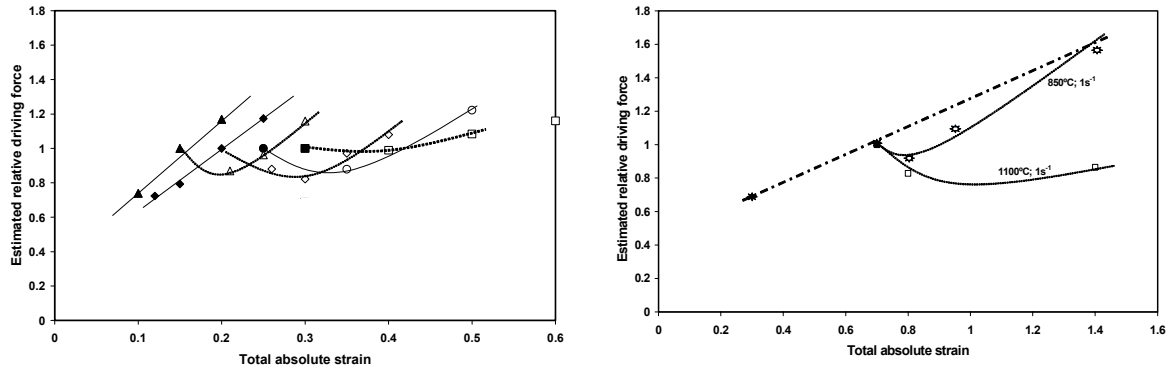


Fig. A2.22: Estimated relative driving pressure, as a function of the total absolute strain. (symbols like in the previous figure).

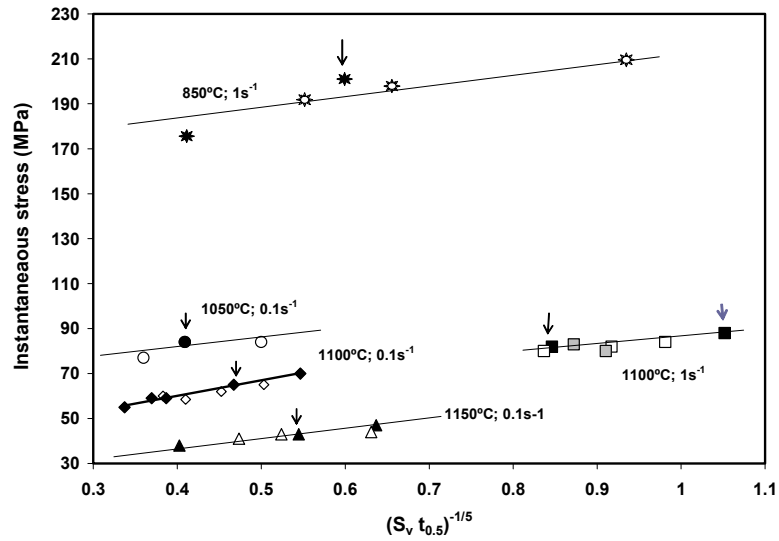


Fig. A2.23: Instantaneous stress before holding, as a function of the product $(S_v t_{0.5})^{-1/5}$. Plain symbols correspond to strain reversal tests

Appendix 3

Detailed Technical Report by Centro Sviluppo Materiali Spa

Authors: J H Bianchi, P Vescovo and F Dionisi Vici

Nomenclature

<i>HWT</i>	Hot Working Theory
<i>H + DRV</i>	Hardening + Dynamic Recovery
<i>DRX</i>	Dynamic Recrystallisation
<i>MDRX</i>	Metadynamic Recrystallisation
<i>SRX</i>	Static Recrystallisation
<i>FEM</i>	Finite Element Method
<i>AR</i>	Area Reduction $\Delta A/A_0$
<i>HDM</i>	Homogeneous Deformation Model
<i>EQR+CVEL</i>	Equivalent Rectangular Areas at Constant Volume Elongation
<i>CL</i>	Constitutive Law
<i>SSRS</i>	Shear Strain Reversal Softening
<i>no-SSRS</i>	Classical Isotropic Hardening model
σ	uniaxial flow stress
$\bar{\sigma}$	equivalent stress from a general stress state
ε	uniaxial strain
$\bar{\varepsilon}$	effective plastic strain from a general strain state
$\dot{\varepsilon}$	effective strain rate
<i>T</i>	temperature
ε_d	critical strain to start DRX
X_d	dynamically recrystallised volume fraction
X_s	statically recrystallised volume fraction
<i>t</i>	time
$t_{0.5}$	time to 50% recrystallisation
<i>D</i>	austenite grain size
$\hat{\varepsilon}$	prior strain rate for SRX and MDRX
$\hat{\varepsilon}_d$	prior deformation critical strain from $\hat{\varepsilon}$
<i>Z</i>	temperature-compensated strain rate
<i>Q</i>	activation energy
s_{def}	hardness variable
s_0	hardness value at yield
s_{ss}	hardness value at steady state
$\bar{\varepsilon}_{pass}$	strain variation in the rolling pass
t_{pass}	total rolling deformation time in a pass
$\langle \dot{\varepsilon} \rangle_{time, strain}$	time and strain weighed average strain rate in a rolling pass
ε_{ch}	strain at which strain rate is changed in stepped strain rate compression testing.
<i>SR</i>	strain rate
<i>rd, td, wd</i>	rolling, thickness and width directions

Summary

The hot working behaviour of two steels of opposite recrystallisation kinetics has been studied by monotonic compression and written as an isotropic hardening viscoplastic constitutive law with a state parameter that evolves with hardening, recovery and recrystallisation. The state parameter hypothesis implies an immediate response of the microstructure to any strain rate change and the simulation results obtained by applying this assumption are compared against corresponding experiments.

As post-deformation recrystallisation in varying strain rate experiments is sensitive to the path history, representative measures of the prior strain rate have been identified in stress relaxation experiments to be used in the process model.

A constitutive softening identified for both steels under shear strain reversal has been used to modify the constitutive law and predictions validated against torsion experiments.

Representative complex loading scenarios in rolling have been identified and analysed by simulations with the resulting Constitutive Law.

Results indicate that FEM hardening rates in compression and torsion testing are about the same, as opposed to classical torsion analysis that gives a larger DRX strain. The best fitting to stress relaxation after varying strain rate deformation is given by the time average prior strain rate. For bar rolling of the medium-C steel the process model predicts DRX at the bar core as opposed to the simplistic model based on area reduction that gives SRX across the whole section. For flat rolling of the non-recrystallising stainless steel and in spite of the differences in shear strain accumulation through passes, the effects on rolling forces between roughing and tandem configurations are minimal, this being due to the predominance of the deformation by rolling elongation. A comparison against corresponding rolling experiments shows in general good agreement in rolling force and power consumption, but the assumed full softening of the yield surface appears as a hypothesis too penalising.

A3.1 Introduction and evaluation of models (WP1)

A3.1.1 CSM objectives

CSM objectives within this project is to better understand the effect of complex loading on rolling variables, on the homogeneity of the rolled product and on tooling forces during hot rolling of:

- a stainless duplex steel (2304) of extremely slow recrystallisation kinetics at the rolling temperatures explored.
- a Carbon-Mn without micro alloying inhibitors (55Cr3) steel, having very fast recrystallisation kinetics

They meant to be representative of the extreme microstructural situations of large grain elongation and early refinement. The work is focused on representative passes of flat and long products rolling schedules.

A3.1.2 State of the art

Classical hot working characterization procedures are based on testing at isothermal, nearly constant strain rate conditions and optimising the fitting to the resulting peak flow stress by using a function of temperature and strain rate. This classical approach delivers a viscoplastic-type of constitutive law that gives a meaningless flow stress with the vanishing of the strain rate, does not incorporate any history along the deformation path and is only suitable for steady state FEM simulation analysis [A3.1-A3.3].

Constitutive models incorporating some of the material history evolution [A3.4-A3.12] were next developed and interfaced to general purposes FEM codes on the assumption the invariants of strain, strain rate and flow stress computed from the general triaxial state in Metal Forming are equivalent to the corresponding measures of the mechanical testing. The early work was concerned with non-recrystallising steels [A3.6-A3.8]. It was later extended to austenite by incorporating post-deformation recrystallisation into interpass models that used different measures of the prior deformation critical strain defining the transition from static recrystallisation (SRX) to metadynamic recrystallisation (MDRX) [A3.1,A3.9-A3.11]. Integrated rolling-interpass models have been applied to determine the effects of rolling parameters on dimensional tolerances and forces and used to derivate simplified models for on-line control [A3.10-A3.11]. Through-process models of complete rolling lines that consider multiple structures [A3.1,A3.9] undergoing different degrees of hardening and volume fractions percolated by grain sizes have been also developed.

The basis of the constitutive formulation above is the assumption of immediate response of the material [A3.8, A3.14] to the changing temperature-strain rate conditions as does occur in the general Metal Forming case, where development of non-uniform distributions of deformation variables are unavoidable. Thus, become important to assess the departure of the FEM prediction based on this hypothesis against experimental testing at changing strain rate, where some delay in the material response might develop.

Previous ECSC funded research found a different mechanical behaviour from monotonic and strain reversing torsion testing [A3.8] of a duplex stainless steel. It was unclear whether the large difference was related to the very slow recrystallising steel then used, to the large percentage of ferrite present at the hot working temperature or to the straining mode. As rolling pass deformation involves also shear strain reversals, a constitutive description that contemplated the material hardening under shear being different than under normal loading was also required to investigate. This requires further investigation for the general three-dimensional case where not all the shear strain components undergo either reversal or suffer the same hardening than the normal components.

Due to the large amount of ferrite at rolling temperatures in the 2304 steel, this soft phase takes most of the strain thus preventing recrystallisation of austenite [A3.7]. For the general case of austenitic steels, recrystallisation during and past deformation needs to be incorporated into the constitutive law and related to the varying strain rate-temperature conditions found in rolling.

The experimental possibilities of following microstructural evolution in industrial mills at intermediate stages are very limited, because of the difficulties in both cutting samples and quenching them in times short enough to freeze the microstructure. The alternative of reproducing industrial rolling conditions in pilot mills is also restricted, because of their smaller geometric dimensions, their slower rolling speeds and the absence of interstand tensions. The presence of DRX and MDRX in the finishing rolling conditions has been

shown in torsion simulation tests and modelled using simple regression equations and spreadsheets [A3.15, A3.16]. Results from numerical simulation rolling models also support this view [A3.17-A3.20], while it is also suggested that the time to trigger DRX cannot be normally exceeded in the rolling pass [A3.21,A3.22].

The characteristics of SRX and MDRX have been discussed by Karjalainen and Perttula [A3.23] and those of MDRX by Hodgson [A3.24], among others. The onset and the kinetics of different softening mechanisms, DRX during deformation and static recrystallisation (SRX) or MDRX after the final pass or during interpass-times are temperature T and strain rate $\dot{\epsilon}$ dependent. Therefore, although the critical strain ϵ_d to start DRX, its associated volume fraction X_d and the post-deformation softening fraction X_s can all be identified almost uniquely (within the experimental error limitations) in isothermal tests at constant strain rates, their extrapolation to the local conditions developed in metal forming is not straightforward.

A more detailed analysis of the problems above requires identification of the complex loading scenarios, definition of a constitutive framework general enough to be used in most of hot rolling situations and validation against experiments.

In this work, a full material characterisation of the two steels under study has been carried out and a constitutive model that considers work hardening, dynamic recovery, dynamic recrystallisation and static softening has been developed [A3.25]. The constitutive model has been experimentally validated for varying $\dot{\epsilon}$ - T conditions, coded in a coupled temperature-stress analysis procedure of the commercial FEM code Abaqus ® [A3.26] and applied to mechanical testing and typical passes of flat and bar rolling. Wherever possible, validation of the numerical predictions against the experiments has been carried out.

A3.1.3 Evaluation of constitutive models

A3.1.3.1 Primary variables

Classical Hot Working Theory (HWT) models only considers the two *primary* variables:

- temperature T
- strain rate $\dot{\epsilon}$

which although well defined in mechanical testing have only a *local meaning* when applied to the *localisations* that any Metal Forming process develops, as revealed by FEM simulations.

These two variables are normally recast in a single ‘Temperature-compensated strain rate’ or Zener-Hollomon variable:

$$Z = \dot{\epsilon} \cdot \exp\left(\frac{Q_{def}}{R T}\right) \quad \dots (A3.1)$$

defined in terms of an activation energy Q_{def} for deformation and $R = 8.31 \text{ J mol}^{-1} \text{ K}^{-1}$.

In the context of Metal Forming FEM analysis, these two *primary variables do not have memory of their own*, their evolution is governed by the global Thermo-Mechanical

incremental equilibrium of the deforming solid, this being driven by the kinematical boundary conditions defined by both the tooling and the symmetry planes and thermal boundary conditions.

A3.1.3.2 Hereditary variables

These variables are related to the intrinsic process variables *history*:

- strain ε
- elapsed time t

and to the

- microstructural evolution of the particular material.

The range of the primary variables is only local to the time t . The range of the hereditary variables is larger, but they tend to fade away simultaneously with the emerging of new ones (i.e.: hardening of recrystallised grains, recrystallisation after subsequent deformation, etc.). As the constitutive model to be developed in this project aims at future applications in through-process modelling, the material description should keep the number of variables propagating throughout the rolling schedule to a minimum.

The simplest constitutive model enriches the classical Hot Working Theory by introducing a '*strength*' or '*hardness*' factor s [A3.4-A3.9, A3.12] that keeps records of the hardening level achieved by the structure at a given stage of hot deformation as result of the competition of hardening and dynamic recovery (H+DRV). The strength and hardness names trace back to the pioneering work of Hart [A3.27], where s was a measure not depending on the primary variables T - $\dot{\varepsilon}$, as it happens in tensile or indentation cold testing.

In order to incorporate the material softening during and past deformation into the constitutive description, the *recrystallised volume fraction* X needs to be incorporated into the framework. To keep the number of variables confined, it is convenient to relate the recrystallised fractions to the single hardness parameter, which then becomes the only carrier of the microstructural evolution through the process [A3.1,A3.9]. Nevertheless, short-range secondary hereditary variables are needed, which are eliminated after their effects are transferred to the hardness carrier. They are:

- the dynamic X_d and static X_s recrystallised volume fractions
- the grain size

The laws defining the material evolution are obtained from experimental data at constant strain rate and temperatures. These equations need to be experimentally validated under varying strain rate conditions and then recast in incremental form to be integrated under the non-isothermal, through-strain rate localization paths that result in rolling.

A3.1.4 Complex loading scenarios during hot rolling

A3.1.4.1 Complex loading *within* a single rolling pass

Representative complex loading scenarios of hot rolling are:

- (a) loading is sensitive to strain rate and temperature (Hot Working Theory) and both these variables are not constant in the deformation zone as in the mechanical test.
- (b) hardening rates for normal and shear deformation might be different.
- (c) dynamic loading resulting from microstructural changes in the steel (DRX).
- (d) dynamic loading as the result of shear strain reversal in the pass.
- (e) uneven straining of the constituent phases [A3.37] in steels that are multiphase at the rolling temperatures.

A3.1.4.2 Complex loading *throughout* rolling passes

In addition to the situations above, representative complex loading scenarios between passes arise from:

- (f) reversal of the two normal strain components associated to the directions transversal to the rolling direction, in long product mills with alternating H-V stands.
- (g) shear deformation *cycles* of different nature in reversing-roughing and tandem-finishing rolling lines.

To the purposes of linking subsequent passes throughout the rolling schedule, the microstructural state also needs to be updated accordingly to the static and post dynamic microstructural changes occurring at the interpass.

A3.1.4.3 Process models

CSM has developed over the years rolling models of different complexity for both flat and bar rolling [A3.2, A3.7-A3.11] using several FEM codes. The models used in this project are based on the commercial Abaqus® [A3.26] code.

To cut down the number of element in the mesh to a minimum, mirroring sections are substituted by boundary conditions on the symmetry planes wherever appropriate.

The stress analysis is performed considering the material is rate and temperature dependent and its history evolves under the constitutive options described.

The thermal solution considers internal conduction within the rollingstock, thermal radiation to the environment, heat transfer to the rolls and heat generation due to conversion of mechanical work both within the stock and at the friction interface.

In addition to the evolution of the unknowns of the system: displacement and temperatures and their derivate fields strain, stresses and thermal cooling/heating rates, a number of auxiliary variables are set up to monitor the onset of DRX and drive its integration,

computation of mean strain rates for SRX/MDRX softening and introduction of softening due to strain reversals. All this features are explained in detail in the section §A3.3 and §A3.4.

One of the main advantages of the FEM analysis is that the material behaviour coded can be applied to a variety of deforming conditions. Thus, the constitutive law finally applied to the validation exercises against rolling results in §A3.5 were previously validated at different stages of its development on several testing conditions ranging from compression, torsion and strain reversals.

A3.2 Materials, process-mechanical testing links, research procedures and partners interacting with CSM work (WP2)

A3.2.1 Materials

Table A3.1 gives the chemical composition of the steels studied in the project.

The 55Cr3 is a medium-C steel, rolled normally to bar and rods for fabrication of springs. It contains no recrystallisation inhibitors. The research on this steel is confined to the stands of the intermediate mill of ORI-MARTIN mill, near the cutting device where routine specimens are taken for inspection.

The 2304 duplex stainless steel is extremely slow in recrystallising, due to the fact that most of the strain during hot rolling is taken by the ferritic phase; it is finally rolled to sheet thickness for cold drawn domestic appliances. The interest in this research is in the THYSSEN-KRUPP ACCIAI SPECIALI TERNI hot rolling line.

A3.2.2 Process- mechanical testing links

The main thematic interconnections between the materials and process selected and the mechanical testing are shown in Table A3.3.

The two steels under study undergo opposite microstructural evolution during and after hot working. The medium-C steel recrystallises very fast [A3.10] both during and after deformation. Dynamic recovery during duplex stainless steel deformation counterbalances hardening giving a perfect steady state stress plateau [A3.7, A3.8] and no recrystallisation takes place after deformation at the standard hot working temperatures and interpass times.

Each of the two materials are associated to different final products and hence hot rolling routes. During duplex stainless steel flat rolling, grain elongation and accumulation of normal strain components throughout passes occur, without any reversal of them. The main shear strain component, however, undergoes a reversal under the roll gap and there is a residual shear strain accumulation through passes at about one quarter of the rolling stock [A3.7].

It is known that the medium-C 55Cr3 steel exhibits very fast recrystallisation kinetics and therefore a significant part of the rolling strain history is expected to be destroyed at interpass with grain refinement and growth overcoming grain elongation. The effective strain might even be reduced within the roll gap as result of dynamic recrystallisation in some passes of the schedule. This scenario is also bound to affect the shear strain reversal in the pass, although such components will not built up through-process accumulation as in the case of the 2304 steel. Those effects are expected to be maximum during the roughing pass, but at the finishing stands some strain accumulation and interaction with static and dynamic recrystallisation might take place. Also, as the result of alternate H-V roll positions,

there is a reversing of normal strain components; depending on the rolling temperatures, such reversing might affect recrystallisation, as found by mechanical testing in a previous ECSC project [A3.8].

Hot working behaviour can be better studied by mechanical testing under controlled temperature strain rate conditions. Although the varying strain rate profiles that develop in rolling process cannot be reproduced experimentally, the mechanical test at constant rate conditions can be used to construct viscoplastic constitutive laws to be used in rolling FEM simulation. This transfer implies to validate first the immediate response of the material to sudden strain rate changes by mechanical testing. Relevant to the current complex loading project, differences in the hardening rate under normal and shear loading need to be investigated by compression and torsion.

The rolling strain rate is a sharply varying instantaneous measure, the history of which is not accounted even in the most sophisticated models. Consequently, prior rolling deformation strain rate measures have to be investigated to make use of the recrystallisation data obtained at constant strain rate mechanical testing.

Shear strain components suffer a reversal under the roll gap, both in flat and bar rolling. In addition, tandem H-V bar rolling configurations introduces a normal elongation-transverse spreading cycle that implies reversing of normal strain components. Mechanical testing represents a flexible tool to understand and quantify the effect of strain reversal on both hardening rate and recrystallisation kinetics.

A3.2.3 Research procedures and CSM-partners interaction

A3.2.3.1 Mechanical testing

A good interaction with the associated partners OULU and CEIT were maintained during the project. The early results were used for development of the constitutive model and the application of this to FEM simulations used for both defining conditions for further testing and validations.

CSM machined specimens of industrially as-cast and rolled material and distributed them to the partners the University of Oulu (55Cr3) and CEIT (55Cr3 and 2304).

For the 55Cr steel mechanically tested by both Centres, CSM took maximum care in cutting the samples from the same non central areas of the cross section, thus minimising the effects of segregation. CSM made a metallographic study on the speed of grain growth using different thermal cycles (Table A3.2). These results were then used to define a common preheating cycle by OULU and CEIT, thus avoiding excessive grain size scattering with a target grain size of 35 μm . It consisted of a heating up to 1050°C, holding 120 s, ramping at 2°C/s to the testing temperature (between 1050 and 950°C), which was next held for 15 s before testing. OULU also used the CSM results in Table A3.2 to generate different austenitic grain size at the deformation temperature in their study of the effect of such parameter on recrystallisation.

The University of Oulu performed monotonic compression, double hit compression, stress relaxation and tension/compression on the medium-C 55Cr3 steel. CEIT performed plane strain compression, and both monotonic and strain reversing torsion on both the 55Cr3 and 2304 steels.

The raw material data were used by CSM to elaborate constitutive laws, accordingly to the general framework detailed in §A3.3. The constitutive laws were then implemented into a general purpose FEM program (§A3.3.3) and used for computing simulations and validation of the mechanical test results (§A3.3.3.1) and representative rolling scenarios (§A3.5).

Important interactions between partners were:

- The University of Oulu validated the deformation response of the steel under *continuous and abruptly changed strain rate* against their results from isothermal *constant* strain rate. The results of this study justifies in a first approximation the use of a state parameter in the constitutive law developed by CSM for the medium-C steel as detailed in §A3.3.1. The FEM-implemented CL was then validated against OULU Gleeble compression and CEIT torsion testing. Of particular interest is the comparison of the results from numerical (immediate response) and experimental varying strain rate testing, which gives a measure of the delay in the response of the real material (§A3.3.3.1.2).
- the recrystallisation kinetics results obtained from stress relaxation testing at constant and abruptly *changed* strain rate by OULU has allowed to calibrate/validate the best prior strain rate measure from general deformation (§A3.3.2.3) to be used in the interstand rolling model.
- the results of the torsion experiments by CEIT have been used to compare torsion and compression hardening rates under monotonic loading conditions (§A3.3.3.1).
- the amplitude of the shear strain reversals obtained from CSM flat and bar rolling simulations (§A3.4.2.1) was used in the definition of the torsion testing programmed by CEIT. Thus, shear strain reversals +0.06/-0.06 for flat rolling of duplex stainless steel and +0.15 / -0.15 for bar rolling of medium-C steel were selected.
- CEIT torsion reversal data for both steels were used to modify the constitutive law for strain reversals and validate the FEM simulation results. (§A3.3.3.1.3).

A3.2.3.2 Pilot mill and industrial hot rolling

The two hot rolling industrial lines of interest associated to the materials in §A3.2.1 are the ORI-MARTIN plant at Brescia and the THYSSEN-KRUPP ACCIAI SPECIALI TERNI.

The ORI-MARTIN mills produce bars of diameters ranging from 18 to 60 mm and wire rods from 42 to 5.5 mm. It uses as-cast feedstock of 140 mm x 140 mm preheated at typical exit temperatures of 1100°C, which are roughed in a single non-reversing rolling stand before entering two intermediate mills that have a total of 13 stands. The line then splits in three finishing mill branches accordingly to the final product: bar, rod or wire. The current research is focused on a stand of the intermediate mills placed before the splitting of the line and after the first loop, where the shear-cutter takes samples of rolled stock for routine checking. This choice allowed characterising the material entering the case study stand and

therefore using a more realistic constitutive behaviour in the FEM simulation. Typical rolling temperatures for the case study stand for the medium-C steel studied are 1030°C- 1040°C with rolling speeds of the order of 1.7 m/s and pass reductions 0.3 -0.45.

The hot line of THYSSEN-KRUPP ACCIAI SPECIALI TERNI rolls slabs of 174 mm thickness to 1.5 to 12 mm. The 2304 steel slabs are preheated to around 1150°C and then rolled first in a reversible roughing stand in 5/7 passes and finally in the 7 stands of the tandem finishing mill. Typical rolling temperatures are 1050-1140°C in the roughing and 920 – 1000°C in the finishing mill. In some cases a 3-roughing passes pre-rolling is used [A3.7]. Typical rolling speeds are 1 to 2 m/s in the roughing and 1.4 to 12 m/s in the finishing, with pass reductions ranging from 0.15 to 0.45.

Representative rolling conditions were studied by rolling experiments at CSM two high stand Danieli, 0.475 m roll diameter pilot mill. Experimental evolution of temperature evolution from unloading, rolling forces and power consumption were recorded through the experiments.

A3.3 Formulation of constitutive laws (WP4), model implementation (WP5) and validation (WP6) against the mechanical testing of WP2

A3.3.1 Viscoplastic constitutive model with one state parameter related to isotropic hardening

The mechanical testing data produced by the University of Oulu and CEIT at constant $\dot{\epsilon}$ - T conditions were used to formulate a rate-dependent, isotropic hardening constitutive law for both materials.

The numerical fitting procedure follows the steps defined by Anand [A3.5], Brown et al [A3.6], modified by Bianchi et al [A3.7, A3.25].

A3.3.1.1 Level I: Hardening + dynamic recovery (2304 and 55Cr3 steels)

Classical Hot Working Theory (HWT) assumes the peak stress represents a balance between hardening and dynamic recovery and can be linked to a ‘Temperature-compensated strain rate’ (Equation (A3.1)). Figure A3.1 shows the medium-C steel experimental peak stress data in a range of temperature strain-rates and their fitting to a function of Z , which allows to write:

$$\bar{\sigma}_{ss} = \bar{s}_{ss} \sinh^{-1} \left[\left(\frac{Z}{A} \right)^{m_s} \right] \quad \dots (A3.2)$$

The fitting for the initial yield stress (Fig. A3.1) gives also a good correlation, but the slope (inverse of the strain rate sensitivity) and s-factor are different than for the peak fitting:

$$\bar{\sigma}_o = s_o \sinh^{-1} \left[\left(\frac{Z}{A} \right)^{m_o} \right] \quad \dots (A3.3)$$

The s-variable has been named ‘hardness’, because of its independence of strain rate and temperature in the early models [A3.27]. This condition not necessarily occurs, as revealed by the different strain rate sensitivities m_o and m_s in Fig. A3.1. Consequently, it is convenient

to take the steady state $\hat{s}_{ss}$ as the reference and normalise the initial hardness accordingly [A3.7, A3.25]:

$$s_o = \hat{s}_o \sinh^{-1} \left[\left(\frac{Z}{A} \right)^{m_o} \right] / \sinh^{-1} \left[\left(\frac{Z}{A} \right)^{m_s} \right] \quad \dots (A3.4)$$

Also, the peak stress does not necessarily represent the hypothetical steady state if any DRX occurs. In such case, the $\hat{s}_{ss}$ value has to be identified from extrapolation of early deformation stresses where DRX is absent [A3.8-A3.9]. The evolution of the material hardness s between the two extremes:

$$s_o \leq s \leq \hat{s}_{ss} \quad \dots (A3.5)$$

evolves accordingly to a constitutive law that needs to be identified. Following previous work [A3.8-11, A3.25], the following equation was explored for the steels under study:

$$\left. \frac{ds_{def}}{d\bar{\varepsilon}} \right|_{H+DRV} = h \sinh^{-1} \left[\left(\frac{Z}{A} \right)^n \right] \left[1 - \left(\frac{s_{def}}{\hat{s}_{ss}} \right)^a \right]^b \quad \dots (A3.6)$$

The flow stress for any intermediate state in Equation (A3.5) is then written with the same functional forms than Equations (A3.2-A3.3) as:

$$\bar{\sigma} = s_{def} \sinh^{-1} \left[\left(\frac{Z}{A} \right)^{m_s} \right] \quad \dots (A3.7)$$

Tables A3.4-A3.5 give the constitutive parameters obtained for the two steels under study.

Curve 2 in Fig. A3.2 shows an example of this constitutive fitting to a 55Cr3 steel experimental data over a strain range up to 0.8. The light decay of the 'steady state' or 'saturation' hypothetical stress associated to $\hat{s}_s$ is due to heat generation.

A3.3.1.2 Level II: Hardening + dynamic recovery + dynamic recrystallisation (55Cr3 steel)

By comparing the Level/constitutive predictions against the experimental data, the evolution of the flow stress softening $\Delta\sigma$ exclusively due to DRX, can be identified (e.g.: Curve 4 in Fig. A3.2) and written in terms of the DRX volume fraction X_d [A3.25] which in turn depends on strain rate, temperature and strain in excess of the DRX strain:

$$\varepsilon_d = B_2 \exp \left[\left(\frac{Q}{RT} \right) \right]^q \dot{\varepsilon}^r \quad \dots (A3.8)$$

Consistently with the formulation of the first level (Equation (A3.6)), the evolution of the two history carriers hardness s_{def} and X_d past the ε_d threshold are also written incrementally:

$$\left. \frac{ds_{def}}{d\bar{\varepsilon}} \right|_{H+DRV+DRX} = \left. \frac{ds_{def}}{d\bar{\varepsilon}} \right|_{H+DRV} - B(Z) \cdot \frac{dX_d}{d\bar{\varepsilon}} \quad \dots (A3.9)$$

$$\frac{dX_d}{d\bar{\varepsilon}} = (1 - X_d) \frac{c}{\varepsilon_d} \frac{k}{\eta} \eta^{k-1} \quad \dots (A3.10(a))$$

if :

$$\eta = \frac{\bar{\varepsilon} - \varepsilon_d}{\bar{\varepsilon}} \geq 0 \quad \dots (A3.10(b))$$

otherwise:

$$\frac{dX_d}{d\bar{\varepsilon}} = 0 \quad \dots (A3.11)$$

$$B(Z) = B_1 \quad Z^p \quad \left/ \sinh^{-1} \left[\left(\frac{Z}{A} \right)^{ms} \right] \right. \quad \dots (A3.12)$$

The total DRX fraction under varying $\dot{\bar{\varepsilon}}$ and T conditions are obtained as:

$$X_d = \int \frac{dX_d}{d\bar{\varepsilon}} d\bar{\varepsilon} \quad \dots (A3.13)$$

Curve 5 in Fig. A3.2 shows the resulting fitting to the experimental softening and Curve 3 the final fit to the flow stress.

The *integral* relationships for the history carriers s_{def} and X_d as obtained from the isothermal-constant- $\dot{\bar{\varepsilon}}$ mechanical testing are not suitable for FEM simulation of hot Metal Forming, because in any of these processes there is material flow through $\dot{\bar{\varepsilon}}$ and T localizations. For these conditions, it is preferred to use *incremental* forms of the history-carriers evolution as given by Equations (A3.6), (A3.9) and (A3.10), and integrate them (e.g.: Equation A3.13) along the material path at the particular quasi-static $\dot{\bar{\varepsilon}}$ and T conditions of each of the small time increments of a FEM analysis. The underlying assumption to this approach is the immediate reaction of the structure to the new local conditions, hypothesis that is discussed against experimental results in §A3.3.3.1.2.

A3.3.2 Post deformation softening: effect of prior-complex loading on recrystallisation kinetics and constitutive law (*Level III*: incorporation of static softening). Case study: 55 cr3 steel

An integral model for post-deformation volume fraction softening X_s was produced by the University of Oulu from double hit and stress relaxation [A3.28] axisymmetric compression:

$$X_s = 1 - \exp [-\beta (t / t_{0.5})^\alpha] \quad \dots (A3.14)$$

$\varepsilon < \varepsilon_d$ Static recrystallisation (SRX):

$$t_{0.5} = 2.75 \cdot 10^{-11} D^{1.06} \varepsilon^{-2.45} \dot{\varepsilon}^{-0.125} e^{184000/RT} \quad \dots (A3.15(a))$$

$\varepsilon > \varepsilon_d$ Metadynamic recrystallisation (MDRX):

$$t_{0.5} = 3.59 \cdot 10^{-2} \dot{\varepsilon}^{-0.785} e^{24800/RT} \quad \dots (A3.15(b))$$

These results require some re-casting in order to be used by the process models, because:

- as opposed to the controlled isothermal conditions of the mechanical testing, the rolling-stock temperature varies in the interstand as consequence of losses by radiation and internal conduction from the hotter inner core to the surface.
- there has not been a unique strain rate (in general also not a unique temperature) during prior rolling deformation.

both issues being examined below. From the resulting X_s for the particular deformation-cooling conditions, the flow stress for subsequent deformation is computed with a *Level III* of the constitutive law using the law of mixtures:

$$s = s_{def} \cdot (1 - X) + \hat{s}_0 \cdot X_s \quad \dots (A3.16)$$

and the base equation (Equation (A3.7)). Available choices to compute the interstand softening X_s in the general Metal forming Case are examined below.

A3.3.2.1 Post-deformation recrystallisation under non-isothermal conditions

The FEM-thermal analysis subsequent to the rolling pass simulation proceeds at small time increments Δt . Taking advantage of this pacing, we can apply quasi-isothermally the Avrami's integral relationship obtained from mechanical testing as described next.

If the volume fraction X_i at the *end* of time t_i is made equal to the *starting* value X_{i+1} of the subsequent increment, an equivalent time t_i^* for the new temperature T_{i+1} results as:

$$t_{i+1}^* = t_i \cdot \left(\frac{t_{0.5}(T_{i+1})}{t_{0.5}(T_i)} \right)^{1/\alpha} \quad \dots (A3.17)$$

and the fraction recrystallised at *the end* of the interval $i+1$:

$$X_{i+1} = 1 - e^{-\frac{\beta}{[t_{0.5}(T_{i+1})]^\alpha} (t_i^* + \Delta t)^\alpha} \quad \dots (A3.18)$$

which allows defining the recursive relationship for the fraction X_n accumulated at time t_n :

$$X_n = 1 - e^{-\beta \left(\sum_{i=1}^n \frac{\Delta t}{t_{0.5}(T_i)} \right)^\alpha} \quad \dots (A3.19)$$

A3.3.2.2 Selecting a prior strain rate from FEM solutions to varying strain rate conditions

The $\dot{\bar{\epsilon}}$ measure influences the reference time $t_{0.5}$ (Equation (A3.15)) and consequently the speed of the recrystallisation kinetics (Equation (A3.14)). Moreover, the threshold for dynamic recrystallisation (Equation (A3.8)) that also delimits the SRX-DRX regimes, depends also on the prior $\dot{\bar{\epsilon}}$.

Several $\dot{\bar{\epsilon}}$ measures have been examined for varying strain rate histories available from FEM solutions:

- a maximum strain rate along the considered deformation path:

$$\dot{\bar{\epsilon}}_{\text{maximum}} \quad \dots (A3.20)$$

- a *strain-averaged* strain rate obtained by weighting the mid-increment *local* equivalent strain rate by its associated fractional equivalent strain increment:

$$\langle \dot{\bar{\epsilon}} \rangle_{\text{strain}} = \frac{\sum \dot{\bar{\epsilon}} \cdot \Delta \epsilon}{\bar{\epsilon}_{\text{pass}}} \quad \dots (A3.21)$$

which might vary across the section; $\bar{\epsilon}_{\text{pass}}$ is the equivalent total strain of the pass computed along the considered material streamline.

- A *time-averaged* strain rate, which also varies across the rolled section, obtained by weighting the mid-increment *local* equivalent strain rate by its associated fractional time increment:

$$\langle \dot{\bar{\epsilon}} \rangle_{\text{time}} = \frac{\sum \dot{\bar{\epsilon}} \cdot \Delta t}{t_{\text{pass}}} \quad \dots (A3.22)$$

where t_{pass} is the total time taken by the deformation.

A3.3.2.3 Assessment of the performance of the FEM prior strain rates measures against stress relaxation experiments

A3.3.2.3 Assessment of the performance of the FEM prior strain rates measures against stress relaxation experiments

In order to assess the most representative strain rate defining the X_s kinetics from prior deformations occurred at varying strain rates, the University of Oulu carried a series of stress relaxation experiments at the temperatures of 950°C and 1050°C on the 55Cr3 steel. CSM applied the prior deformation measures described in §A3.3.2.2 above, to the OULU mechanical testing conditions.

In accordance with typical rolling pass conditions, all OULU tests were carried up to total strains of about 0.5 to 0.6 and at abruptly changed $\dot{\varepsilon}$ conditions. The recrystallised volume fraction X_s was compared against the Avrami fittings (Equation (A3.14)) corresponding to constant $\dot{\varepsilon}$'s (Fig. A3.3). Their results point to a shift of the recrystallisation Kinetics to the conditions defined by the second strain rate.

From the experiments in which the strain rate was changed from 10 s⁻¹ to 1 s⁻¹, the prior strain rate measures computed by Equations (A3.21) and (A3.22) are:

$$\langle \dot{\varepsilon} \rangle_{strain} = 9 \text{ s}^{-1} \quad \text{and} \quad \langle \dot{\varepsilon} \rangle_{time} = 1.5 \text{ s}^{-1}$$

For 1050°C, the X_s Avrami fittings (Equation (A3.14)) obtained using these rates in the $t_{0.5}$ computations (Equation A3.15) show that at the highest strain rate the best result is given by the strain-average strain rate initially, but for longer times the best performance is the time-averaged strain rate. For the testing at 950°C under the same conditions [A3.25], and at 1050°C for the slower strain rate levels (Fig. A3.3), the best results are obtained by using the time average. The time-average procedure, i.e. the mean strain rate, was also found reasonable by Karjalainen et al [A3.31] for a Ti-bearing steel, especially in the case of SRX and decreasing strain rate. They observed that the 'effective strain rate' was close to the final (second) strain rate.

A3.3.3 Implementation of constitutive laws in a FEM code and validation against mechanical test results

The material law obtained from the deformation conditions under Gleeble axisymmetric compression was next implemented as rate-dependent plasticity into the commercial FEM code Abaqus®.

The FEM simulations make the assumption that the second invariant of the stress and strain components (the equivalent stress and strain) do not depend on the complexity of the deformation state. As this varies in the several situations considered: compression, torsion, bar and flat rolling, it is important to validate the simulations against the experimental results and assess whether corrections to the basic constitutive behaviour obtained in §A3.3.1 and §A3.3.2 are required.

A3.3.3.1 Constitutive law FEM validation of compression and torsion experimental data. Case study: 55 Cr3 steel

A3.3.3.1.1 Monotonic loading

Figure A3.4 shows the application of the FEM model to simulate a Gleeble compression test at 1050°C and 1 s^{-1} , revealing the excellent agreement between the raw experimental axial load data obtained by OULU and the simulation by CSM. It is interesting to note also the uniformity of the plastic strain distribution up to plastic strains around 0.6.

Next, the model was applied to a torsion experiment from CEIT under the same nominal hot working conditions. Figure A3.5(a) shows a detail of the simulating conditions, in which the symmetry plane at mid-gauge length has been exploited to cut down computing.

The resulting plastic strain distribution reproduces well its linear dependencies across the specimen radius (Fig. A3.5(b)) and amount of twist θ (proportional to the time at constant twist rate conditions in Fig. A3.5(c)) assumed by both the Fields-Backoffen [A3.29] and effective radius [A3.30] classical analyses. The linearity of the strain with the twist(time) breaks down at the specimen shoulder from strains about 0.33 (Fig. A3.5(c), strain rate= 1 s^{-1}). In general, this local breakdown develops only on a very narrow band that is about 4% of the specimen half-gauge length (Fig. A3.5(d)).

Figure A3.6(a) shows the evolution of the thermal field at different surface locations along the specimen length. The results have been obtained on the assumptions of a constant initial temperature 1050°C, internal conduction, adiabatic heating with a 0.9 conversion factor of mechanical work into heat, zero flux towards the shoulder, no radiation losses but also no electric heating. The higher temperature localization towards the middle of the gauge length (Fig. A3.6(c)) justifies the specimen break up there, result that is not available in the conventional Fields-Backoffen model. The thermal evolution in time is shown in Figure A3.6(b).

A comparison of the raw torque-time experimental and simulation results is shown in Fig. A3.7. The agreement is very good in the hardening-dynamic recovery region and up to early dynamic recrystallisation but the FEM predictions start to deteriorate at strains beyond 0.7. This is adequate for the bar rolling schedule of the steel under consideration, where the maximum pass strain is no larger than 0.8 and is completely erased in the following interpass due to the fast MDRX recrystallisation kinetics. Two possible reasons can explain the mismatch at larger strains in Fig. A3.7. The first is the breakdown of the thermal conditions imposed in the simulation to follow the real conditions of the experiment. The second is that the numerical predictions were computed with a constant grain size that in the current simulation version does not evolve as DRX progresses. Should the mean grain size be updated by incorporating the DRX grain refinement, this is bound to lower the critical strain to DRX and hence to accelerate the softening, thus bringing the simulation torque closer to the observed experimental values.

Figure A3.8 shows a comparison of the resulting stresses obtained by the different methods. The experimental stress level from the Gleeble compression data is about the same that the one obtained from the raw torque curve using the conventional classical model [A3.29-A3.30] based on the product of a power of strain times power of strain rate. However, the strain to peak stress-hence the threshold for DRX- predicted by the torsion classical model is larger than both the current model and the compression experiments.

It can also be seen in Fig. A3.8 how the Gleeble experimental strain-stress data are very well matched by the FEM local stresses in torsion at the middle of its specimen gauge length. Even the stress at the torsion specimen shoulder does not exhibit a large departure from the compression-testing curve.

A3.3.3.1.2 Varying strain rate loading

The University of Oulu next carried out experiments in which the constant strain rate was abruptly changed at a certain strain ε_{ch} under both increasing and decreasing conditions simulating rolling. The experiments were done with ε_{ch} occurring at both the hardening-dynamic recovery region and past the onset of DRX (Figs. A3.9 and A3.10). A comparison of the results against the constant strain rate testing showed that within the experimental scatter the flow stress of the second $\dot{\varepsilon}$ regime tends to approach asymptotically the single- $\dot{\varepsilon}$ curve [A3.25]. There is always a short transient that is the composite effect of the strain at which the applied $\dot{\varepsilon}$ stabilises at its new value, dependent on the response of the Gleeble simulator, and a purely material response lagging behind the mechanical state [A3.31,A3.32].

CSM constitutive model in §A3.3.1 is based on the instantaneous response of the structure to the new strain rate conditions. Thus, a comparison of the FEM (instantaneous) and experimental (real material) results under the same kinematics condition gives an estimate of the delay in the material response.

The application of the FEM simulation §A3.3.3 to the experiments, enforcing the same kinematical conditions by using exactly their ramping displacements, avoids any Gleeble machine dynamic effect in the simulation. The results in Figs. A3.9 and A3.10 show in general terms a reasonable agreement. Some scattering is apparent at the highest strain rate due to the experimental difficulty in maintaining this constant. The difference is bigger near the yield point, where even the improvement given by Equation (A3.4) fails to fit properly. The agreement, however, in general tends to improve on reaching stress levels near saturation. The clear effect of recrystallisation on the stress levels after strain rate changes can also be seen.

For all increasing and decreasing strain rates the FEM (immediate response) predictions comes earlier than in the real material. The delay is of the order of 0.02 in strain.

A3.3.3.1.3 Strain reversals (*Constitutive Model Level IV*)

Both in flat and bar working stocks, the rolling process involves a sharp and fast shear strain reversal under the roll gap (Figs. A3.14 and A3.15) and a subsequent inversion at the next stand if interstand microstructural changes do not erase the history (§A3.4.3 and §A3.4.4). The set of torsion experiments agreed with CEIT reflected that amount of shear strain and inversion times, with a sequence composed of immediate reversal followed by reversal after a holding time of about 1 s.

CEIT torsion experimental results clearly showed a lower flow stress level on the loading subsequent to the first reversal [CEIT Appendix in this Report], respect to the monotonic case. It is important to note that such behaviour not only appeared in the extremely fast recrystallisation non-micro alloyed medium-C steel in the austenite region currently studied, but was also previously detected in a non-recrystallising duplex stainless steel [A3.8, A3.35].

The current constitutive model so far described is based on a flow stress that relies on an ever-increasing equivalent plastic strain (Equation (A3.26)) that does not react to reversals. Consequently, in order to simulate the experimental results within this constitutive framework, the resulting hardness from Equations (A3.6) and (A3.9) has to be softened after the first reversal.

Two hypothesis were examined and implemented in the FEM code:

(a) anisotropy creep *yield surface* [A3.26,A3.36]. defined as:

$$\tilde{\sigma} = \left[F (\sigma_{22} - \sigma_{33})^2 + G (\sigma_{33} - \sigma_{11})^2 + H (\sigma_{11} - \sigma_{22})^2 + 2L \sigma_{23}^2 + 2M \sigma_{31}^2 + 2N \sigma_{12}^2 \right]^{0.5} \quad \dots (A3.23(a))$$

where the coefficients $F, \dots, N$ are function of '*potential ratios*' defined in terms of stresses of single deformation modes (i.e.: compression, torsion, etc.) as:

$$R_{ij} = \frac{\sigma_{ij}}{\tilde{\sigma}} \text{ for } i=j \quad R_{ij} = \sqrt{3} \frac{\sigma_{ij}}{\tilde{\sigma}} \text{ for } i \neq j \quad \dots (A3.23(b))$$

The effects of shear strain γ reversals were then simulated by modifying the R_{ij} as follows:

$$\begin{aligned} R_{ij} &= 1 \text{ for any monotonic loading} \\ R_{ij} &= 1 \text{ for } \gamma < \gamma_{\text{first reversal}} \\ R_{ij} &= 0.8 \text{ for } \gamma > \gamma_{\text{first reversal}} \end{aligned} \quad \dots (A3.24)$$

In this hypothesis *only* the shear stress component that underwent strain reversal was intended to be reduced. The early FEM results produced with this hypothesis for the torsion reversal experiments were not conclusive and showed little difference respect to the conventional isotropic hardening constitutive law that evolves in terms of the ever increasing plastic strain.

(b) rate dependent isotropic plasticity with the whole yield surface softened after the first reversal as:

$$s_{def} = 0.8 \hat{s}_{ss} \quad \text{for } \gamma > \gamma_{\text{first reversal}} \quad \dots (A3.25)$$

The simulation attempts by using Equations (A3.24)-(A3.25) to introduce a Shear Strain Reversal Softening (SSRS) of the flow stress in the constitutive law should be regarded as an approximation to the real complex loading situation and considers a flat stress level for any strain beyond the first reversal, with elimination of the hardening rate.

A comparison of the experimental results against FEM predictions using the monotonic loading constitutive model (Equations (A3.6) and (A3.9) for the history carrier and the basic Equation (A3.7) for the stress) and the new level accounting for the softening due to reversals (Equation (A3.25)) is shown in Fig. A3.11. Note that torque has been plotted as its absolute value to compare changes of level after reversal. The agreement is good enough to consider the constitutive model validated for the torsion conditions.

A3.4 Constitutive laws for complex loading (WP4) and application to rolling (WP5)

A3.4.1 Equivalent strain definitions used in multiaxial loading

Practically all current FEM work on process modelling uses constitutive laws obtained from the stress/strain data resulting under the nearly uniform strain state achieved in mechanical testing and this is carried at constant strain rate $\dot{\varepsilon}$ and temperature T . These conditions contrast with rolling deformation, where several deformation modes might become activated at varying $\dot{\varepsilon}$ and T .

Two plastic strain measures are found in FEM codes. They are the equivalent strain:

$$\bar{\varepsilon} = \int \dot{\varepsilon} dt \quad \dots (A3.26)$$

where $\dot{\varepsilon}$ is the always-positive strain rate defined from its components as:

$$\dot{\varepsilon} = \left[\left(\frac{2}{3} \right) \dot{\varepsilon}_{ij} : \dot{\varepsilon}_{ij} \right]^{1/2} \quad \dots (A3.27)$$

and the effective strain:

$$\bar{\varepsilon} = \left[\left(\frac{2}{3} \right) \varepsilon_{ij} : \varepsilon_{ij} \right]^{1/2} \quad \dots (A3.28)$$

updated from the ε_{ij} strain components, these not necessarily increasing monotonically. The last condition is particularly true for the shear strain components developed towards the rolling stock border, which change sign as revealed by the very well known 'Neutral Point' of flat rolling.

The measures given by Equation (A3.26) and Equation (A3.28) only coincide under proportional loading conditions, which is not always the case of the complex loading situation that develops in metal forming.

A3.4.2 FEM models of rolling

The constitutive law-FEM model developed and validated in §A3.3 was next applied to rolling case studies. The first stage §A3.4.2.1 - §A3.4.2.2 of this application implies the underlying hypotheses:

- in presence of localizations, the material reacts instantaneously to any new $\dot{\varepsilon}$ - T conditions, therefore following an equation of state.
- the constitutive law is invariant in terms of the equivalent total strain in spite of the differing deformation state found in the tests and in the rolling pass.

While a second stage §A3.4.3-§A3.4.4 considers:

- incorporation of Shear Strain Reversal Softening (SSRS) effects.

Tables A3.6 and A3.7 show typical conditions used in the FEM simulations of both bar and flat rolling.

A3.4.2.1 Main process variables affecting the constitutive behaviour

Temperature is the process variable of more impact to the constitutive behaviour. Figure A3.12 shows the temperature distribution predicted by the FEM analysis for the medium-C steel at the 8th stand of the bar mill studied and detail of its evolution in time during the subsequent interpass. In spite of the fast cooling of the rolling stock surface when contacting the rolls, its temperature recovers within the interpass due to conduction from the internal core. For the conditions studied in Table A3.6, the temperature difference between core and surface at the entry of the subsequent stand is about 90°C, with a 25°C maximum temperature rise due to conversion of mechanical work.

The process variable of second constitutive importance is the strain rate. Figure A3.13 shows examples of strain rate distributions predicted by the model for the stand under study. The pattern of the effective strain rate normally used in the constitutive description is shown in Fig. A3.13(a) and the contribution of its main normal and shear components in Fig. 13(b and c) respectively; the meaningful values of these last justify a constitutive characterisation by both compression and torsion. Even though the shear strain in the pass is relatively low (Figs. A3.14-A3.15), its reversal entails a high shear strain rate change (-11 to $+12 \text{ s}^{-1}$ in Fig. A3.13(c)). As the strain rate vanishes outside the roll gap, a meaningful value is required to compute recrystallisation kinetics; Fig. A3.13(d) shows the average measure computed accordingly to §A3.3.2.2.

The process variable of third constitutive importance is the strain, which is related to the rolling reduction. Figure A3.14 shows the evolution of the strain components as revealed by a simulation of a typical flat rolling pass of duplex stainless steel 2304, at three through-thickness locations: (a) midplane, (b) a quarter thickness and (c) underneath stock skin. The results have been produced for a rolling reduction $r=25.8\%$, corresponding to the engineering strain 0.264 for plane strain conditions, using the base constitutive law *Level I* described in §A3.3.1.1. It can be clearly seen that the rolling-thickness directions (rd, td) shear strain component near the strip symmetry mid-plane tends to zero but its absolute amount increases towards the rollingstock surface with a very well defined reversal. For this case study the amplitude of the reversal is from -0.052 to $+0.044$ at quarter thickness and from -0.07 to $+0.064$ at a point 1.05 mm under the contact skin.

Bar-rolling simulation results for a typical round-oval pass, for an engineering area reduction $AR=38 \%$ are shown in Fig. A3.15. The simulation has been produced with the *Level II* constitutive law for 55Cr3 medium-C steel introduced in §A3.3.1.2 in which the hardness is modified to incorporate DRX. As in the flat rolling case, the shear strain components towards the midplane of the bar parallel to the roll axis (points (a) and (d)) are very small. Their value increase towards the border in contact with the rolls (points (b) and (c), located about 2 mm under the skin of the rolled bar). The shear component defined by the pair (rd, td) always undergoes some reversal in the range $+0.25$ to -0.08 (point (b)) and $+0.15/-0.14$ (point (c)). The component (rd, wd) undergoes a non-monotonic shear straining but without significant reversion. Near the inflection point of the grooved roll (c) there is also a significant shear strain component (td, wd).

The simulation results for both flat and bar rolling in Fig. A3.14 and Fig. A3.15 clearly show that as a consequence of the shear strain reversal at the outer layers of the stock, the more

realistic effective strain computed with Equation (A3.28) is smaller than the equivalent value obtained from Equation A3.26.

Another important result of the FEM simulation is their difference respect to the conventional area reduction AR industrial parameter for bar rolling. While in the case of flat rolling the AR=25.8% overall parameter is close to the 0.27-0.29 equivalent strains (Fig. A3.14), in the case of bar rolling an AR=38% implies equivalent strains around 0.7 (Fig. A3.15).

A3.4.2.2 Recrystallisation. Case Study: medium-C steel 55Cr3, round-oval pass

A3.4.2.2.1 Dynamic recrystallisation

Figure A3.16 shows the evolution of the DRX volume fraction X_d predicted by Equation A3.13, which is obtained by adding up Equation A3.10 contributions each increment the strain, exceeds the instantaneous ε_d given by Equation (A3.8). The maximum X_d occurs around the centre streamline (I) and decreases towards the contacting surface (II) (Fig. A3.16). The last result is due to the increase of ε_d (Equation (A3.8)) mainly as consequence of the T -drop produced by the chilling effects of the rolls (Fig. A3.12(c)), combined with the lower $\bar{\varepsilon}$ resulting from the reversal of its shear strain components towards the bar border (Fig. A13.15). Figure A3.16 also indicates that the bar's outer layers will undergo SRX as consequence of $X_d=0$ while the core will suffer MDRX.

A3.4.2.2.2 Interstand recrystallisation

Interstand softening influences the material strength and therefore the demand to the tooling at the subsequent stand. Two main issues needs to be considered:

A3.4.2.2.2.1 Representative strain rate from previous rolling deformation

The strain rate influences both the critical strain for dynamic recrystallisation ε_d (Equation (A3.8)) and determines whether the subsequent softening is static or metadynamic (Equation (A3.15)). As rolling deformation involves material flow with varying strain rate, a representative strain rate needs to be defined.

The simplest Homogeneous Deformation Model (*HDM*) considers that the whole volume of the bar under the rolling gap was subjected to a mean, constant strain rate obtained as a ratio of the Area Reduction divided by the roll-gap residence time. A more realistic model uses a strain rate obtained from Equivalent Rectangular areas, Constant Volume Elongation and residence time *HDM(EQR+CVEL)* [A3.33 – A3.34]. As the strain rates computed by FEM models vary locally and are much higher than those predicted by simplified models, the alternative measures introduced in §A3.3.2 are now applied to rolling.

A comparison of the recrystallisation predictions for different prior strain rate definitions for the stand under study is shown in Table A3.8 and discussed next:

The over simplistic *HDM(AR)* predicts:

$$\varepsilon_{HDM} = 0.27 < \varepsilon_d(\dot{\varepsilon}_{HDM}) = 0.31 \quad (\text{SRX}) \quad \dots \quad (\text{A3.29})$$

while the most realistic *HDM(EQRCVEL)* gives:

$$\varepsilon_{HDM} = 0.55 > \varepsilon_d(\dot{\varepsilon}_{HDM(EQR+CVEL)}) = 0.38 \quad (\text{MDRX}) \quad \dots (\text{A3.30})$$

results those are constant across the product section.

On the contrary, the measures introduced in §A3.3.2 vary across the section. Their computation along streamlines of the FEM solution, for the two typical exit locations in Fig. (A3.16), gives for the maximum strain rate along the path:

$$\text{bar core: } \bar{\varepsilon}_{FEM} = 0.73 > \varepsilon_{d \max SR} = 0.43 \quad (\text{MDRX}) \quad \dots (\text{A3.31(a)})$$

$$\text{bar border: } \bar{\varepsilon}_{FEM} = 0.37 < \varepsilon_{d \max SR} = 0.44 \quad (\text{SRX}) \quad \dots (\text{A3.31(b)})$$

and for the average prior strain rate measures:

$$\text{bar core: } \bar{\varepsilon}_{FEM} = 0.73 > \varepsilon_{d FEM \text{ averages}} = 0.38 \text{ to } 0.39 \quad (\text{MDRX}) \quad \dots (\text{A3.32(a)})$$

$$\text{bar border: } \bar{\varepsilon}_{FEM} = 0.37 > \varepsilon_{d FEM \text{ averages}} = 0.32 \text{ to } 0.30 \quad (\text{MDRX}) \quad \dots (\text{A3.32(b)})$$

Although the FEM average strain rate measures gives larger strain and critical recrystallisation strain than the HDM-EQRCVEL model, both of them predict MDRX along the considered streamlines. The FEM averages along path, however, are preferred because of the validation exercise described in §A3.3.2.3 (Fig. A3.3).

A3.4.2.2.2 Interstand recrystallisation under non-isothermal conditions

The recrystallised volume fraction X at the interpass is traditionally computed in terms of the elapsed time and the local temperature, assumption that overestimates the recrystallisation Kinetics. Consistently with the treatment of all the other history variables in the constitutive framework of this research, the Avrami equation is also written *incrementally* and integrated under the non-isothermal conditions determined by the interpass cooling as detailed for the general case in §A3.3.2.1.

Figure A3.17(a-c) shows the amount of FEM pass-strain in excess of the instantaneous dynamic strain determined with some of the $\dot{\varepsilon}$ definitions above, for the 55Cr3 medium C steel and rolling pass subsequent to the microstructural changes in (Fig. A3.16), with an initial interstand temperature 1040°C. No substantial differences are seen, result which is due to the predominance of the temperature factor in Equation (A3.15) and the low exponent affecting the strain rate in ε_d Equation (A3.8), Table A3.4).

Recrystallisation is very fast at the end of the following interstand (Fig. A3.17(d)) and no significant differences in volume fraction X_5 results by applying the various models [A3.11], although the scatter in grain sizes from SRX and MRDX can be significant. Two regions develop slower recrystallisation kinetics: a fraction of the contacting layers cooled by the rolls and a zone at the free surface; although this last remains at higher temperature than the former, has underwent lower strain in the previous rolling pass.

The effects of the different strain rate hypotheses on the rolling forces are almost negligible because of the extension in volume of the fully recrystallised area and the small product cross section. The torque, however, which depends on the material strength at the border, shows some minimal sensitivity to the choice of the strain rate (Fig. A3.17(e)).

A3.4.3 Strain reversal effects within a single rolling pass. Case study: medium-C steel 55Cr3, oval-round pass

A3.4.3.1 Strain reversal effects on the local strain, flow stress and forces

The different constitutive response of the material to single straining modes examined in §A3.3.3.1 needs next to be mapped upon the hot rolling case, where many of these modes coexist (Fig. A3.14 and Fig. A3.15). The constitutive model validation against mechanical testing has shown that for *monotonic* loading the constitutive law from the axial *compression* Gleeble experiments gives also good approximation to *shear strains* up to about 0.8 (Figs. A3.7 and A3.8). The next issue to consider is how to incorporate the effect of *shear strain reversals* into the rolling model.

Of the two hypothesis examined in connection with the torsion twist reversal, and in view of the results validated in Fig. A3.11, the second of them given by Equation (A3.25) was selected with

$$\gamma = \varepsilon_{rd,td} \quad \dots (A3.33)$$

where *rd* and *td* are the main rolling and thickness direction respectively. The switching during the FEM rolling simulation from the conventional hardness resulting from Equations (A3.6), (A3.9) and (A3.16) to the softened hardness in Equation (A3.25) is driven by an auxiliary variable that monitors the change in sign of the shear strain component given by Equation (A3.33).

Figure A3.19 compares the strain distribution across a rolled bar section for the conventional and modified constitutive law obtained for the 55Cr3 medium-C steel. As the result of the material softening due to reversal (Fig. A3.18) the second model predicts a bigger plastic strain on the same area (Fig. A3.19).

The modified SSRS constitutive law predicts a stress softening at the bar border not found in the conventional isotropic model (Fig. A3.20), which in turn affects the demand to the tooling. The results in Fig. A3.21 for an intermediate rollingstock exiting at 52 mm diameter, show that the differences in rolling forces are almost negligible, but there is 12% difference in the predicted torque. For the wider rolling stocks resulting in flat rolling, both forces and power consumption show difference to the incorporation of the strain reversal softening effect in the constitutive law (§A3.5.2).

A3.4.3.2 Influence of shear strain reversals on DRX

The differences between the equivalent (Equation (A3.26)) and the strain-reversal-influenced effective plastic strain (Equation (A3.28)), shown in Fig. A3.14 and Fig. A3.15, are also reflected in the pace of the dynamic recrystallisation kinetics. For the bar rolling pass of the medium C-steel examined, DRX is maximum towards the centre of the stock but the difference resulting from both strain definitions is minimal due to the small shear strains (Fig. A3.22, point (a)).

Although DRX at the outer layers of the rolling stock (e.g. point (b)) is of lower magnitude because of the chilling effects of the rolls raising the threshold ε_d , the influence of the shear strain reversal on both effective strain and recrystallised volume fraction is significant.

A3.4.3.3 Influence of DRX on shear strain reversals

There is a marked difference on the effect of the strain reversal on the two materials under study. The duplex stainless steel does not suffer DRX and therefore the outer layers of the strip undergo in each rolling pass a full shear strain reversal (Fig. A3.14(b and c)). On the other hand, the very fast recrystallisation non-micro alloyed medium-C steel develops DRX within the deformation zone reducing the magnitude of the strain that propagates. On the assumption that all strain components undergo the same softening than the effective strain:

$$\{\varepsilon_{ij}, \bar{\varepsilon}\} = (1-X_d) \cdot \{\varepsilon_{ij}, \bar{\varepsilon}\} \quad \dots (A3.34)$$

unrecovered strains smaller than the kinematics predictions of *Level I* will result.

Figure A3.23 shows the reducing effect of DRX on the strains predicted by the FEM model. The presence of substantial DRX at the bar core has a significant effect on the effective strain (point a), but its effect on the small *shear* components there is minimal.

Because DRX diminishes towards the bar border (*II*) in Fig. A3.16 and point (b) in Fig. A3.22), its effects on the *unrecovered strains* also diminish (Fig. A3.23, point (b)) respect to the core. In particular, there is almost no effect on the shear strain. The *unrecovered shear strains* at the bar outer layers are therefore significant and might preset reversals at the entry of the next rolling stand if the SRX/MDRX in the following interpass is not complete.

A3.4.4 Strain reversal effects through rolling passes

A3.4.4.1 Through passes shear strain reversal during flat rolling of 2304 steel in reversing and tandem configurations

As the result of its negligible interstand recrystallisation [A3.7-A3.8,A3.35], the 2304 steel is ideal to study the effects of strain accumulation and residual shear strain after reversal. The torsion reversal tests performed by CEIT for the 2304 steel [A3.7-A3.8, A3.35 and current report] showed a Shear Strain Reversal Softening (SSRS) that has been incorporated into the constitutive model as a full contraction of the yield surface (Equation (A3.25)) once a the shear strain (Equation (A3.33)) first reverses.

The constitutive model defined by the CL-*Levels I* (§A3.3.1.1) and *IV* (§A3.3.3.1.3) was next applied to FEM simulation of both roughing and tandem configurations. Figure A3.24 shows the areas affected by shear strain reversal during two consecutive passes. The extension of the SSRS area through thickness does not present significant differences between the two configurations, with the only exception of a band at the midplane.

The evolution of the residual shear strain at different thickness is shown in Fig. A3.25. It is clear that for the same reductions, the reversing stands deliver a stock with larger shear strain accumulation and this builds up further through passes.

Figure A3.26 shows FEM predictions of specific rolling forces and power consumption per unit width for the roughing and tandem configurations, under the classical (*NO- SSRS*) and *SSRS* constitutive assumptions at equal initial temperature and pass reduction. As in the matching experiments (§A3.5), the simulation assumes that the microstructure in the first pass has been homogenized during the reheating, eliminating any cumulative effects. It can be seen that the difference between the *NO-SSRS* and *SSRS* predictions in the second pass is significant. At equal temperature, rolling speed and reductions, both the load and

power demanded by the second pass appears to be slightly lower in the case of roughing sequences than in tandem sequences (Fig. A3.26).

A3.4.4.2 Through-passes strain reversal of normal strain components in H-V bar mill configurations. Effect of interstand recrystallisation (medium-C steel)

A case study to assess the amount of normal strain reversal taking place between the H-8th and V-9th stands of ORI-MARTIN bar mill was carried out. At the H-8th stand exit, the effective strain $\bar{\epsilon}$ distribution across the oval section ranges from 0.4 to 0.73 (Fig. A3.15), while the normal strain component ϵ_{33} in the rolling direction exhibits a reduced scatter from +0.35 to +0.38. The normal strain components in the direction *transversal to rolling*, however, show more variation. Thus, the ϵ_{11} component varies from -0.15 at the free surface to -0.73 at the core (Fig. A3.27 (a)), the negative sign indicating fiber compression across the whole bar section. Full recrystallisation ($X=1$) at the 8th –9th interpass destroys the previous strain history, with the net result of an ϵ_{11} at the 9th stand varying from a – 0.20 fiber compression at the tool/free surface edge to a +0.71 fiber elongation at the core (Fig. A3.27(b)). If recrystallisation is incomplete, however, some reversal will take place under the roll gap at the 9th stand. Assuming that all strain components undergo the same softening than the equivalent strain, the accumulated strain component ϵ_{11}^{acc} at the $(l+1)$ th stand can be written as:

$$\epsilon_{11}^{acc} = (1-X) \cdot \left. \epsilon_{11} \right|^{stand(l)} + \left. \epsilon_{11} \right|^{stand(l+1)} \dots (A3.35)$$

In addition to this strain reduction due to recrystallisation, the normal components ϵ_{11} and ϵ_{22} will undergo additional reduction due to their reversal. In the extreme case of recrystallisation absence ($X=0$), the resulting ϵ_{11}^{acc} varies from +0.11 elongation near the symmetry midplane to –0.5 shortening at the tool/free surface border (Fig. A3.24(c)). Although this variation is dependent on the groove geometries, the general trend is a reduction of the spread of the transversal-to-rolling-direction normal strains due to the alternating H-V configuration.

The general incomplete recrystallisation case ($0 < X < 1$) will fall between the two extreme situations shown in Fig. A3.24(b and c). Thus, Fig. A3.24(d) shows the ϵ_{11}^{acc} computed for the X fraction in Fig. A3.17.

For the particular 55Cr3 steel and rolling conditions examined, recrystallisation is complete almost elsewhere and therefore the ϵ_{11} and ϵ_{22} strains will not either propagate or undergo significant reversal during deformation in the next rolling stand. This scenario is not bound to change in the following mill stands because of the temperature raise due to mechanical working and the effect of such variable on the $t_{0.5}$ (Equation A3.15).

An important result from Fig. A3.24 is that there is more straining *under* the second roll gap in the directions transversal to rolling when previous interpass recrystallisation has been full. Strain reversals due to incomplete recrystallisation in presence of H-V stands tends to cancel out strain accumulation on those directions. The rolling-thickness directions *shear*

strain also undergoes reversal under the roll pass (Fig. A3.15), and this is also sensitive to the magnitude of interstand tensions.

A3.5 Investigation of constitutive behaviour by experimental (WP2) and simulation (WP6) rolling. validation

A3.5.1 Experimental rolling

Plates of both 2304 duplex stainless steel and 55Cr3 medium-c steel were rolled in the 470 mm diameter, two-high Danieli mill at CSM. The 35-40 mm thick specimens were reheated at the temperatures detailed in Table A3.9 for 1 hour. Accordingly to the experimental testing carried by CEIT and OULU, the selected rolling temperature and reductions define non-recrystallising conditions for the 2304 steel and full interpass recrystallisation in the 55Cr3 steel. thus, the experiments on the 2304 could be used to investigate cumulative effects between passes, whilst the rolling on the 55Cr3 defines almost independent passes.

Continuous data were acquired from the mill load cells at intervals of 0.05 s and from a thermocouple embedded at midthickness of the plate at intervals of 0.5 s. Figure A3.28 shows an example of the temperature records for the case study A30 in Table A3.9.

Both the tension and current motor consumption were recorded at the much higher frequency of 8192 acquisitions per second (Fig. A3.29(a and b)). These data were later decoded using a Fourier Transform algorithm which eliminates the contributions of frequencies greater than the first significant noise frequency and returns the data thus filtered. Figure A3.29(c and d) shows typical experimental rolling force and power consumption data obtained for the two passes of case study A30.

A3.5.2 Application of the fem simulations of rolling and validation

The experimental results were next used to validate the predictions of the fem model. For the first pass of the non-recrystallising 2304 steel and for all independent passes of the fast recrystallisation 55Cr3 steel, the predicted rolling forces were within a -4% to + 14% error for the 2304 steel and +8% to +18% for the 55Cr3 steel (Table A3.9). The associated power consumption is in an error band respect to experiments of -7% to +9% (2304) and -11% to +6% (55Cr3).

For the 2304 steel, the absence of recrystallisation makes the effect of the strain reversal in the *first* pass to influence the *second* pass. Figs. A3.27, A3.30 and A3.31, show that both the separating force and the power consumption for the second pass differ depending on whether the SSRS or NO-SSRS constitutive hypothesis is used. The resulting error of the force FEM predictions, was between -24% and +7% (one exception at +11%) for the SSRS hypothesis and -9% and 12% (one exception at +19%) for the NO-SSRS. The associated error in power was -10% to +16% (with one exception at +27%) for the SSRS and 0 % to 19% (exception at +33%) for the NO-SSRS. The two main causes of errors are: the slowdown of the roll speed when the mill is loaded and the definition of a steady state with the short rollingstock length used (300 mm).

The numerical predictions for force and power under both the SSRS and classical NO-SSRS hypothesis are compared against the experimental data for one case study in Figs. A3.30 and A3.31. The analysis is restricted to the second pass, where any softening due to strain reversal could be detected and both roughing and tandem configuration are examined. Being practically impossible to attain the same temperature in all the experiments, all

computations have been converted to the temperature of the roughing R2 pass, which is then taken as a reference, by using

$$\{Power\ or\ Force\}_{compensated\ to\ T_{R2}} = CF \cdot \{Power\ or\ force\}_{at\ real\ temperature\ T_{T2}} \quad \dots (A3.36)$$

where:

$$CF = \operatorname{arcsinh}\left\{ \left[\frac{(Z_{(T_{R2})})}{A} \right]^m / \left[\frac{(Z_{(T_{T2})})}{A} \right]^m \right\} \quad \dots (A3.37)$$

and the Sims' mean strain rate [A3.38] was used to compute the Zener-Hollomon parameter Z (Equation (A3.1)).

Both the experimental and FEM predictions (Figs. A3.30-A3.31) indicate that once this conversion has been made, the power/force demanded by a reversing pass is only marginally lower than for the equivalent conditions tandem pass.

Figure A3.32 compares the measured power at the 8th stand of ORI-MARTIN mill against the FEM torque prediction, this converted to power using the nominal speed and a transmission efficiency overall factor 0.85.

A3.6 Final discussion and conclusions

The research in this project aimed at identifying a general framework for development of constitutive equations for representative complex loading scenarios. constitutive laws incorporating not only the reaction to the instantaneous variables strain rate and Temperature -as normally assumed in classical hot working theories- but also including the material history have been developed for two steels of extremely different recrystallisation kinetics. The resulting constitutive law have been implemented into a fem code for thermal-stress analysis and extensively applied and to mechanical testing and rolling cases and the results validated against corresponding experiments.

The main outcomes of the project for the bar rolling of the 55Cr3 steel are:

- Hardening rates from compression and torsion are comparable. Monotonic torsion results have been reproduced by a FEM simulation using a constitutive law from monotonic compression up to strains of 0.7.
- The stress levels, hardening/softening behaviour and location of peak strains of both compression and torsion modes predicted by the FEM model are about the same up to strains 0.7. This result contrasts with the classical torsion analysis, which predicts about the same stress level but larger DRX strains.
- The steel nearly follows an equation of state in presence of strain rate changes, with a strain delay about 0.02.
- Shear strain reversal in torsion introduces a material softening that has been implemented within the isotropic hardening constitutive framework as a weakening of the whole yield surface.
- The effects of different prior strain rate measures on the interpass recrystallisation have been examined. In spite of the large scattering in these

$\dot{\varepsilon}$ values for the rolling pass studied, the effect on rolling forces is minimal because of the massive recrystallised volume. There is however, a minimum effect on the torque because its dependency on the steel strength at the non-recrystallised outer layer.

- The medium-C spring steel exhibits very fast recrystallisation kinetics. For the round-oval, 8th pass of the intermediate mill considered, the hypothesis of immediate reaction of the microstructure to the instantaneous $\dot{\varepsilon}$, T conditions leads to DRX in a wide central area of the bar. Recrystallisation at the end of the following interpass is nearly complete apart from two small zones at the rolled and free surfaces. Both the FEM model and the second HDM model (EQRCVEL) with compatible constant volume elongation predict DRX and MDRX, in contrast with the over simplistic first HDM model based only on area reduction that predicts SRX.
- The reversal of the normal strain components in the direction perpendicular to rolling during bar rolling is significant, but its effect for the considered H-V stands is minimised by interstand recrystallisation. There is more normal straining in the directions transversal to rolling when recrystallisation in the previous interpass has been complete.
- The influence of the shear strain reversal on DRX is higher towards the rolling stock surface than at its centre, lowering there the DRX volume fraction predicted by the equivalent strain.
- Because of its distribution on the bar section, DRX has a minimal effect on reducing the value of the shear strain towards the workpiece surface contacting the rolls. Thus, in general most of the shear strain that underwent reversal in a pass will reverse again at the next stand. However, in the case of the fast recrystallisation medium-C studied, SRX/MDRX will virtually erase previous strain during the interpass, thus preventing any further reversal.

The main outcomes of the project for the flat rolling of the 2304 steel are:

- In spite of the differences in the through-passes accumulation of reversing shear strain under reversing and tandem rolling configurations, the extension of the area affected by shear strain reversal through thickness is almost the same.
- In contrast with the bar mill, the rolling stock width in flat rolling enhances differences in both forces and power consumption in the 2nd pass. When all conditions are converted to the same temperature, the differences between tandem and reversing configurations are however negligible.
- A comparison of experimental and simulation rolling results shows that the hypothesis of full softening of the yield surface after the shear strain reversal is too penalising. The experiments values are bounded between these SSRS predictions and the one predicted by the *Level 1* that does not consider strain reversal effects.

A common result to both steels and process are:

- CEIT reported a development of material softening due to shear strain reversal. This feature has been consequently incorporated into both steel constitutive descriptions with the results for the particular conditions reported above.
- The delay in the microstructural response resulting from comparison of the instantaneous numerical solution and the real material response might have a strong impact on DRX during bar rolling, where the residence time is very short. For a typical 5 ms residence time, the results in §A3.3.3.1.2 suggest that the delay in strain rate adjustment might take place even within the gap [A3.25].

A3.7 References

- [A3.1] Bianchi, J H: 'From Computational Models of Hot Deformation to Process Design and Control', *4th World Congress on Computational Mechanics*, Buenos Aires, S. Idelsohn, E. Oñate y E. Dvorkin eds., CIMNE publ, Barcelona (1998).
- [A3.2] Mirabile, M, Bianchi, J, Buenten, R, Buessler, P, Ingham, P, Mamalis, M, Monfort, G, Requejo, F and Turpel, P: 'Application of FEM to Hot Rolling and Deep Drawing', *Technical Steel Research Final Report N° EUR 15803*, Office for Official Publ. of the E.U., Luxembourg (1997).
- [A3.3] Montmitonet, P and Buessler, P: 'A review of the theoretical analysis of rolling in Europe', *ISIJ Int.*, 31 (1991), pp525-538.
- [A3.4] Perdrix, C: 'Caracteristiques d'écoulement plastique du metal', *Technical Steel Research Final Report EUR 8697FR*, Official Publications of the EU, Luxembourg (1987).
- [A3.5] Anand, L: 'Constitutive equations for rate dependent deformation of metals'. *ASME J. of Eng. Mater. and Technol.*, 104 (1982), pp12-17.
- [A3.6] Brown, S B, Kim, K and Anand, L: 'An internal variable constitutive model for hot working of metals'. *International Journal of Plasticity*, 5 (2), (1989), pp95-130.
- [A3.7] Bianchi, J H and Urcola, J J: 'Development of a model for hot strip rolling of stainless duplex and non oriented magnetic steels', *Technical Steel Research Final Report EUR 18378*, Official Publications of the EU, Luxembourg (1998).
- [A3.8] Zhou, M, Wiklund, O, Karjalainen, L P, Bianchi, J H, Gutierrez, I and Peura, P: 'The effect of strain reversal and strain-time path on constitutive relationships for metal rolling/forming processes', *Technical Steel Research Final Report EUR 19891*, Official Publications of the EU, Luxembourg (2001).
- [A3.9] Bianchi, J H: 'Coupling of microstructural rate effects with the Mechanics of rolling deformation', in: *Proc. 5th Int. Conf. on Computational Plasticity*, Barcelona, Cimne press, D.R.J. Owen et al. ed. (1997), pp1276-1283.
- [A3.10] Thorp, J H, Bianchi, J H, Dahm, B, Mehren, D, Johnson, C P and Dionisi Vici, F: 'Development of systems to aid flexible rolling of long products to closer tolerances by on-line calculation of loaded roll gap, control of stand speed and reheat'. *Technical Steel Research Final Report EUR 20473*, Official Publications of the EU, Luxembourg (2002).

- [A3.11] Bianchi, J H and Dionisi, F: 'Effects of temperature, rolling speed and preset variations on the loaded roll gap response of a bar mill stand', *Proc. 14th Rolling Conference*, The IAS, Buenos Aires, C. Genzano ed. (2002), pp103-112.
- [A3.12] Estrin, Y and Mecking, H: 'A unified phenomenological description of work hardening and creep based on one-parameter models', *Act. Met.*, (1984), 32(1), pp57-70.
- [A3.13] Bianchi, J H: 'A viscoplastic constitutive law with internal variables for Multipass Hot Rolling', *Proc. 4th Int. Conf. on Computational Plasticity*, Barcelona, Pineridge Press, D.R.J. Owen et al. eds. (1995), pp2283-2292.
- [A3.14] Urcola, J J and Sellars, C M: 'Effect of Changing Strain Rate on Stress-Strain Behaviour during High Temperature Deformation', *Acta Metal*, 1987, 35, pp2637-2647.
- [A3.15] Maccagno, T M and Jonas, J J: 'Correcting the effects of static and metadynamic recrystallisation during the laboratory simulation of rod rolling', *ISIJ Int.*, (1994), 34 (7), pp607-614.
- [A3.16] Maccagno, T M, Jonas, J J and Hodgson, P D: 'Spreadsheet modelling of grain size evolution during rod rolling', *ISIJ Int.*, (1996), 36(6), pp720-728.
- [A3.17] Choi, S, Lee, Y, Hodgson, P and Woo, J S: 'Feasibility study of partial recrystallisation in multi-pass hot deformation and application to calculation of mean flow stress', *J. Mater. Process. Technol.*, (2002), 125-126, pp63-71.
- [A3.18] Lee, Y, Choi, S and Hodgson, P D: 'Integrated model for thermo-mechanical controlled process in rod (or bar) rolling', *J. Mater. Process. Technol.*, (2002), 125-126, pp678-688.
- [A3.19] Serajzadeh, S, Mirbagheri, H and Karimi Taheri, A: 'Modelling the temperature distribution and microstructural changes during hot rolling of low carbon steel', *J. Mater. Process. Technol.*, (2002), 125-126, pp89-96.
- [A3.20] Anelli, E: 'Application of mathematical modelling to hot rolling and controlled cooling of wire rods and bars', *ISIJ Int.*, (1992), 32(3), pp440-449.
- [A3.21] De Ardo, A J: 'Fundamental aspects of the Physical Metallurgy of Thermomechanical Processing of Steels', in: *Proc. Int. Conf. on Physical Metallurgy of Thermomechanical Processing of Steels and Other Metals, THERMEC-88*, Tokyo, the ISIJ, I. Tamura ed. (1988), pp20-29.
- [A3.22] DeArdo, A J: 'Modelling of deformation processing: wonderful tool or wishful thinking', in: *Proc. Int. Symp. Mathematical Modelling of Hot Rolling of Steel*, Metallurgical Society of The Canadian Institute of Mining and Metallurgy, Montreal, S. Yue ed. (1990), pp220-238.
- [A3.23] Karjalainen, L P and Perttula, J: 'Characteristics of static and metadynamic recrystallization and strain accumulation in hot-deformed austenite as revealed by the stress relaxation method', *ISIJ Int.*, (1996), 36 (6), pp729 - 736.

- [A3.24] Hodgson, P D: 'The Metadynamic Recrystallisation of Steels', Int. Conf. Thermomechanical Processing of Steels and Other Materials, *THERMEC-97*, The Minerals, Metals and Materials Society, T. Chandra et al. eds (1997), pp121-131.
- [A3.25] Bianchi, J H and Karjalainen, L P: 'Modelling of dynamic and metadynamic recrystallisation during bar rolling of a medium carbon spring steel', *Journal Materials Processing Technologies*, Elsevier, (2005), Vol. 160, No. 3, pp267-277.
- [A3.26] HKS, ABAQUS *User's manual* (2001-2004).
- [A3.27] Hart, E W: 'Constitutive relations for the non-elastic deformation of metals', *ASME Journal of Eng. Mat. and Tech.*, (1976), 98, pp193-202.
- [A3.28] Karjalainen, L P: 'Stress relaxation method for investigation of softening kinetics in hot deformed steels', *Mater. Sci. Technol.*, (1995), 11,(6), pp557-565.
- [A3.29] Pöhlandt, K and Tekkaya, A: 'Torsion testing plastic deformation to high strain rates', *Mat. Sc. and Tech.*, Nov. 1985, 1, pp973-977.
- [A3.30] Fields, D and Backoffen, W: 'Determination of the strain hardening characteristics by torsion testing', *Proc. ASTM*, 1957, 57, pp1259-1272.
- [A3.31] Karjalainen, L P, Porter, D and Peura, P: 'Recrystallization of Ti-microalloyed steels under constant and varying deformation conditions'. Proc. 4th Intern. Conf. on Recrystallization and Related Phenomena, *Proceedings of ReX'99 (JIMIS-10)*, (1999), pp721-726.
- [A3.32] Karjalainen, L P, Somani, M C, Peura, P and Porter, D: 'Effects of strain rate changes and strain path on flow stress and recrystallization kinetics in Nb-bearing Microalloyed Steels', *Proc. Int. Conf. Thermomechanical Processing of Steels, London, IOM*, (2000), pp130-139.
- [A3.33] Lee, Y: 'Calculating model of mean strain in rod rolling process', *ISIJ Int.*, (1999), 39 (9), pp961-964.
- [A3.34] Lee, Y, Kim, H J and Hwang, S M: 'Analytic model for the prediction of mean effective strain in rod rolling process', *J. Mater. Process. Technol.*, (2001), 114, pp129-138.
- [A3.35] Iza-Mendia, A, Piñol-Juez, A, Urcola, J J and Gutierrez, I: Microstructural and mechanical Behaviour of a Duplex stainless Steel under Hot working Conditions, *Metallurgical and Materials Transactions*, 1998, Vol. 29A, pp2975-2986.
- [A3.36] Hill, R: *Proc. Roy. Soc., A* 193 (1948), pp281-310.

- [A3.37] Sellger, R, Agren, J, Hillert, M, Hoglund, L and Bianchi, J: 'Control of phase transformation during processing of partially bainitic multi-component strip steels for controlling the work-hardening characteristics', *Technical Steel Research, Final Report N° EUR 20582*, Office for Official Publ. of the E.U., Luxembourg (2003).
- [A3.38] Sims, R B: "The calculation of roll force and roll torque in hot rolling mills", *Proc. Inst. Mech. Eng.*, (1954), 168, pp191-200.

Table A3.1: Chemical composition of the steels under study

2304 duplex stainless steel								
Cr	Ni	Mo	Mn	Si	P	S ppm	C	N
23.1	4.83	0.22	1.3	0.51	0.024	10	0.025	0.095

Medium-C 55Cr3 steel										
C	S	P	Mn	Si	Cu	Sn	Cr	Ni	Mo	Al
0.55	0.002	0.007	0.78	0.31	0.10	0.005	0.76	0.07	0.02	0.004

Table A3.2 Thermal treatment trials used to characterise the grain growth kinetics.
R.T.= room temperature; W.Q.= water quenching.

Test	T_1 (°C)	T_2 (°C)	t_1 (s)	t_2 (s)	$D(\mu\text{m})$
B1	900	1150	60	120	40
B2	900	1150	60	300	56
B3	900	1050	60	120	34
B4	900	1150	120	120	49
B5	900	-	60	-	16

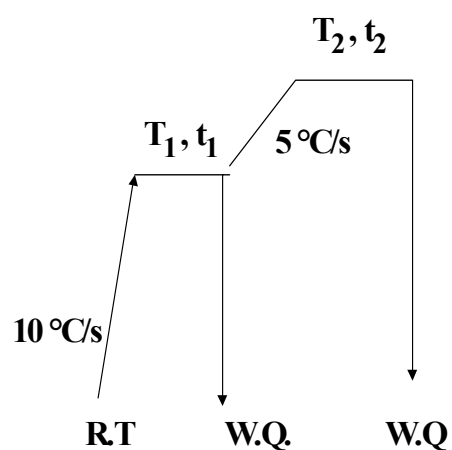


Table A3.3 Process-Mechanical Testing links of CSM work.

MATERIAL		⇐ No RX			Very fast RX ⇒	
PROCESS	Duplex stainless steel 2304		Medium-C steel 55Cr3			
	FLAT ROLLING		BAR ROLLING			
Rolling		- Grain elongation and normal strains accumulating throughout passes - Shear strain reversal - Residual shear strain accumulating through passes at about ¼ thickness (function of interstand tension)	↑ roughing finishing ↓	- Grain refinement and growth destroying strain history and grain elongation - Interaction DRX/strain - RD-thickness shear strain component reversing within the pass/interaction with DRX - Some amount of grain elongation and normal strains accumulating through passes/ interaction with DRX and previous interpass RX - Normal strain reversing between H-V passes. Interaction with DRX and previous interpass RX		
Mechanical Testing	Monotonic compression, tension and torsion	- Viscoplastic (Zener Hollomon) hot working constitutive law - Investigate differences in strain hardening during normal and shear straining				
	Strain rate changes in compression and torsion	- Influence of prior strain rate on RX - Validity of immediate response (eq. of state) assumption				
	Strain reversals			- Shear strain reversal - H-V stands normal strain reversal		
	- Torsion - Compression/tension - Influence on RX	Shear strain reversal				
		Strain rate reversals: - Delay RX kinetics - Lower flow stress				

Table A3.4: Parameters of the isotropic hardening viscoplastic constitutive law for the medium-C 55Cr steel

Q	340169	J. mol ⁻¹
A	3.1613 · 10 ¹²	s ⁻¹
R	8.31	J. mol ⁻¹ · K ⁻¹
m _S	0.231	
m ₀	0.293	
n	-0.02	
$\hat{s}_S$	90	MPa
$\hat{s}_0$	40	MPa
h	1100	MPa
a	1	
b	1	
c	0.5583	
d	0.5	
k	1.5723	
p	0.084	
q	0.103	
r	0.182	
B ₁	1.025	MPa
B ₂	9.597 · 10 ⁻³	
β	0.693	
α	1.5	

Table A3.5: Parameters of the isotropic hardening viscoplastic constitutive law for the 2304 duplex stainless steel

Parameter	Value	Units
Q	550000	J. mol ⁻¹
A	2.11 · 10 ²⁰	s ⁻¹
R	8.31	J. mol ⁻¹ · K ⁻¹
m _S	0.223	
m ₀	0.223	
n	0.01	
$\hat{s}_S$	65	MPa
$\hat{s}_0$	45	MPa
h	445	MPa
a	0.7	
b	1	

Table A3.6: Typical parameters of a simulation for a round-oval rolling pass, intermediate mill, medium-C steel

Process parameter	Symbol	Value
(a) Rolling stand entry		
Bar diameter	ϕ_0	0.052 [m]
Initial Temperature	T_i	1040 [°C]
(b) Stand		
Roll diameter	$\emptyset$	0.56 [m]
Pass reduction	$\Delta A/A_0$	0.3-0.5
Rolling speed	Ω	7.3 [rad s ⁻¹]
Friction coefficient, rolling direction	μ_{RD}	0.35
Friction coefficient, transverse direction	μ_{TD}	0.35
Roll temperature	T_r	150 [°C]
Bulk inelastic fraction converted into heat	β_r	0.9
Contacting surface:		
-Frictional sliding energy fraction converted into heat	β_s	1.0
-Partition between contacting surfaces of generated heat between rolls and bar	β_{rs}	0.5
Interstand travelling time ahead of the stand	$t_{i+1} - t_i$	1.04 [s]
(c) General thermal parameters		
Air temperature	T_a	20 [°C]
Emissivity	ε_T	0.78
Radiation to air shape factor	F_s	1.0
Coefficient of heat conduction to the rolls	h_b	$h_b = a+b \cdot (1-e^{-0.03 p})$ [W m ⁻² K ⁻¹] $a = 10000$, $b = 47000$, p =contacting pressure [MPa]
Thermal expansion coefficient	α_{exp}	$1.48 \cdot 10^{-5}$ [°C ⁻¹]
Conductivity	K	27.2 [W °C ⁻¹ m ⁻¹]
Specific heat	C_s	632 [J Kg ⁻¹ °C ⁻¹]
Density	ρ	7510 [kg m ⁻³]

Table A3.7: Typical parameters of a simulation for a flat rolling pass, 2304 steel;
IM = finishing stand of industrial mill, PM = pilot plant mill

PROCESS PARAMETER	SYMBOL	VALUE
(a) Rolling stand entry		
Slab thickness	H_0	0.040 (IM) – 0.035 (PM) [m]
Initial temperature	T_i	1050 – 1100 [°C]
(b) Stand		
Roll diameter	$\emptyset$	0.71 (IM) – 0.47(PM) [m]
Pass reduction	$\Delta H/H_0$	0.20 - 0.40
Rolling speed	Ω	5.5 (IM) - 4 (PM) [rad s ⁻¹]
Friction coefficient, rolling direction	μ_{RD}	0.35
Friction coefficient, transverse direction	μ_{TD}	0.35
Roll temperature	T_r	200 (IM) – 10 (PM) [°C]
Bulk inelastic fraction converted into heat	β_r	0.9
Contacting surface:		
-Frictional sliding energy fraction converted into heat	β_s	1.0
-Partition between contacting surfaces of generated heat between rolls and bar	β_{rs}	0.5
Interstand travelling time ahead of the stand	$t_{i+1} - t_i$	0.6 - 0.8 (IM) [s]
(c) General thermal parameters		
Air temperature	T_a	20 [°C]
Emissivity	ε_T	0.85
Radiation to air shape factor	F_s	1.0
Heat conduction to the rolls	h_b	$h_b = a+b \cdot (1-e^{-0.03 p})$ [W m ⁻² K ⁻¹] $a=10000$, $b=47000$, p =contacting pressure [MPa]
Thermal expansion coefficient	α_{exp}	$1.39 \cdot 10^{-5}$ [°C ⁻¹]
Conductivity	K	28 [W °C ⁻¹ m ⁻¹]
Specific heat	c_s	661 [J kg ⁻¹ °C ⁻¹]
Density	ρ	7430 [kg m ⁻³]

Table A3.8: Comparison of dynamic recrystallisation strain for different prior strain rate definitions for the streamlines exiting the roll gap at positions (I) and (II) in Fig. A3.16

Strain rate definition	Centre streamline				Border streamline			
	Strain rate (s ⁻¹)	$\dot{\epsilon}_d$	Strain	RX prediction	Strain Rate (s ⁻¹)	$\dot{\epsilon}_d$	Strain	RX prediction
$\dot{\epsilon}_{HDM}$	5.2	0.31	$\epsilon_{HDM}=0.27$	SRX	5.2	0.31	$\epsilon_{HDM}=0.27$	SRX
$\dot{\epsilon}_{HDM(EQR+CVEL)}$	11.5	0.38	$\epsilon_{HDM}=0.55$	MDRX	11.5	0.38	$\epsilon_{HDM}=0.55$	MDRX
$\dot{\epsilon}_{strain-averaged}$	18.6	0.39	$\bar{\epsilon}_{FEM}=0.73$	MDRX	11.6	0.32	$\bar{\epsilon}_{FEM}=0.37$	MDRX
$\dot{\epsilon}_{time-averaged}$	15.5	0.38		MDRX	8.3	0.30		MDRX
$\dot{\epsilon}_{max}$	25.3	0.43		MDRX	28.7	0.44		MDRX

Table A3.9: Comparison of experimental and FEM simulation results. Flat rolling in the 470 mm diameter two high Danieli mill at CSM.

ID	Furnace temperature (°C)	Rolling temperature, core (°C)	Entry strip height (mm)	Height reduction	Rolling force				Power consumption			
					(Tons)		Error (%)		(kW)		Error (%)	
					Experi 1	FEM* (no SSRS) 2	FEM* (SSRS) 1	(NO SSRS) 2	Experi. 1	FEM* (no SSRS) 2	FEM* (SSRS) 1	(NO SSRS) 2
2304 DUPLEX STAINLESS STEEL												
initial width=100 mm												
A15-1	1050	996	35	0.4	107	108	108	1	1	213	218	-3
A15-R2		899	21	0.4	146	147	125	1	-14	223	212	-4
A13-1	1050	1003	35	0.4	102	106	106	3	3	185	213	9
A13-R2		912	21	0.4	146	141	119	-3	-18	230	204	-10
A16-1	1060	1003	35	0.4	99	106	106	6	6	199	213	1
A16-T2		925	21	0.4	142	135	116	-5	-18	225	201	-10
A-29-1	1060	1003	35	0.4	102	106	106	3	3	202	213	-1
A-29-T2		926	21	0.4	147	133	116	-9	-21	225	200	-10
A-30-1	1060	970	35	0.5	133	139	139	5	5	275	305	5
A-30-R2		900	17.5	0.5	184	168	141	-9	-24	267	241	-9
A9	1060	1003	35	0.1	38	39	39	1	1	53	44	-7
A32-1	1240	1150	35	0.3	46	44	44	-4	-4	92	81	-2
A32-R2		1000	24.5	0.3	74	88	81	19	11	108	136	27
A34-1	1020	937	35	0.3	112	112	112	0	0	200	197	-7
A34-R2		863	24.5	0.3	126	141	129	11	2	192	214	13
A35-1	1020	957	35	0.3	91	104	104	14	14	179	185	-3
A35-T2		878	24.5	0.3	120	134	128	12	7	182	209	16
55Cr3 MEDIUM-C STEEL												
initial width=103 mm												
B5-1	1080	1050	35	0.3	80	86	86	8	8	168	158	-11
B5-R2		974	24.5	0.3	86	102	101	18	17	157	166	3
B7-1	1180	1110	35	0.3	64	71	71	11	11	125	131	-1
B7-R2		1035	24.5	0.3	74	85	85	16	15	129	140	6

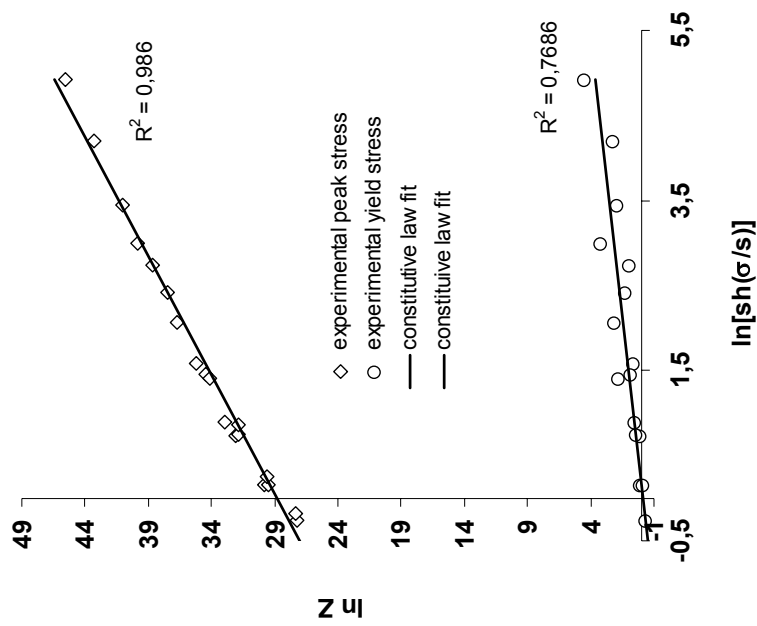


Fig. A3.1: Yield and peak stresses in term of the temperature compensated strain rate. Comparison of the constitutive fitting against monotonic axisymmetric compression Gleeble data for medium C steel 55Cr3 from the University of Oulu.

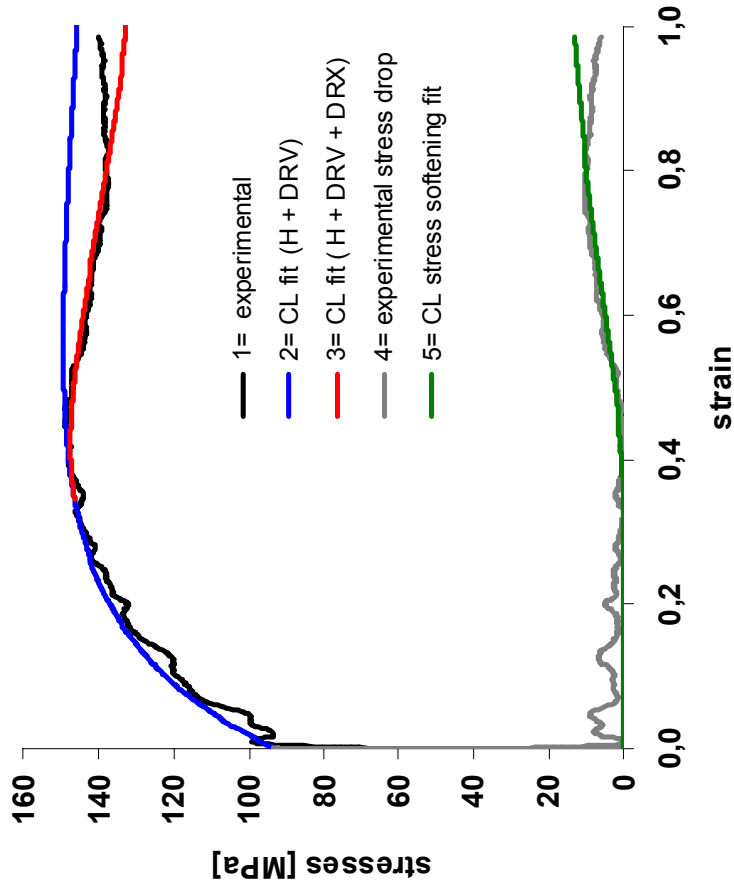


Fig. A3.2 History effects during monotonic loading considered by the proposed constitutive model constructed from Oulu's Gleebe compression data. H=hardening, DRV=dynamic recovery, DRX=dynamic recrystallisation. Medium-C 55Cr3 steel tested at 1050°C, 1 s⁻¹.

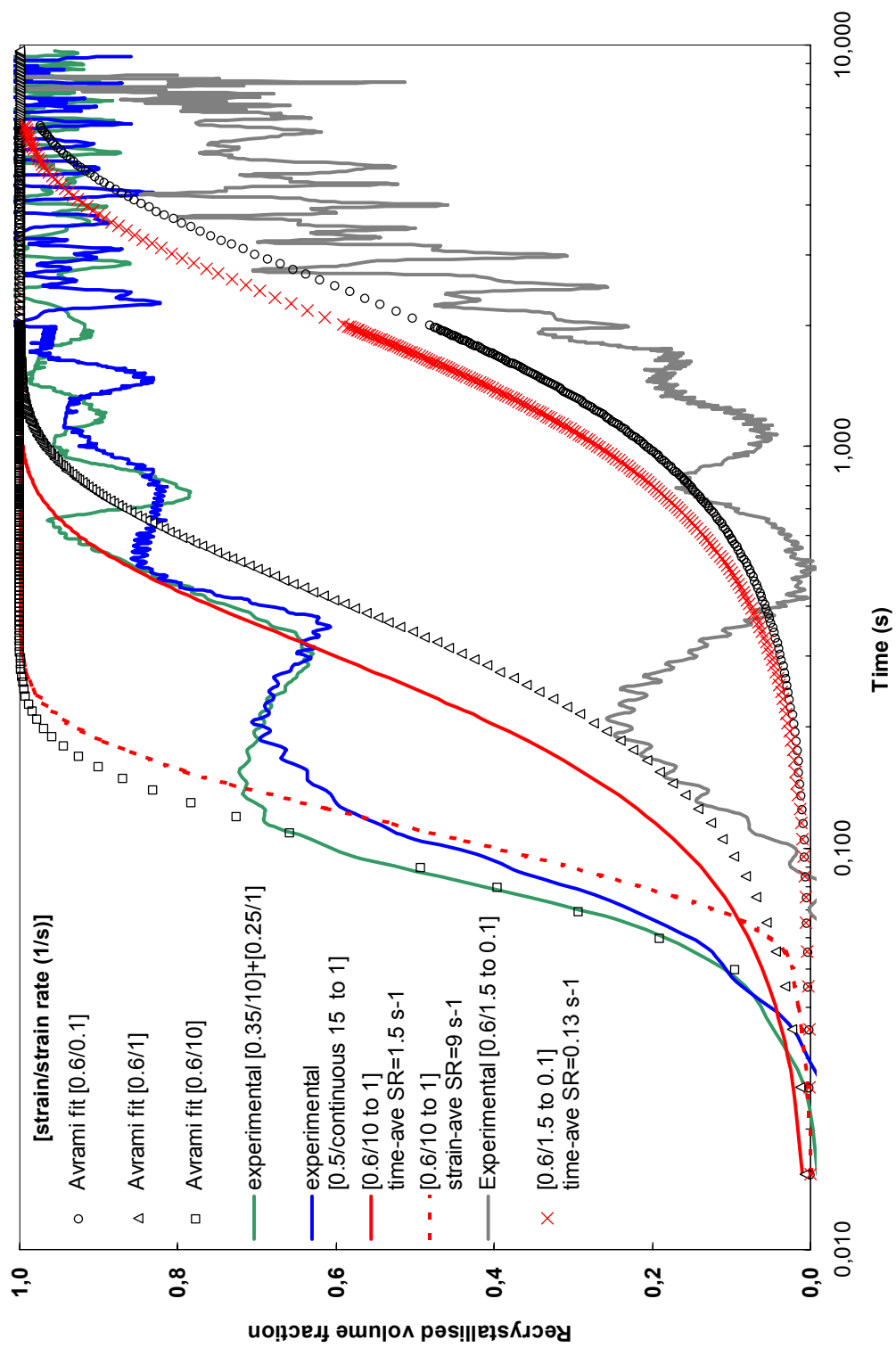


Fig. A3.3: Effect of prior strain rate on static softening. Comparison of Oulu experimental results from constant and changing strain rate with CSM FEM model mean strain rate measures. Stress relaxation experiments at 1050°C on the medium-C 55Cr3 steel.

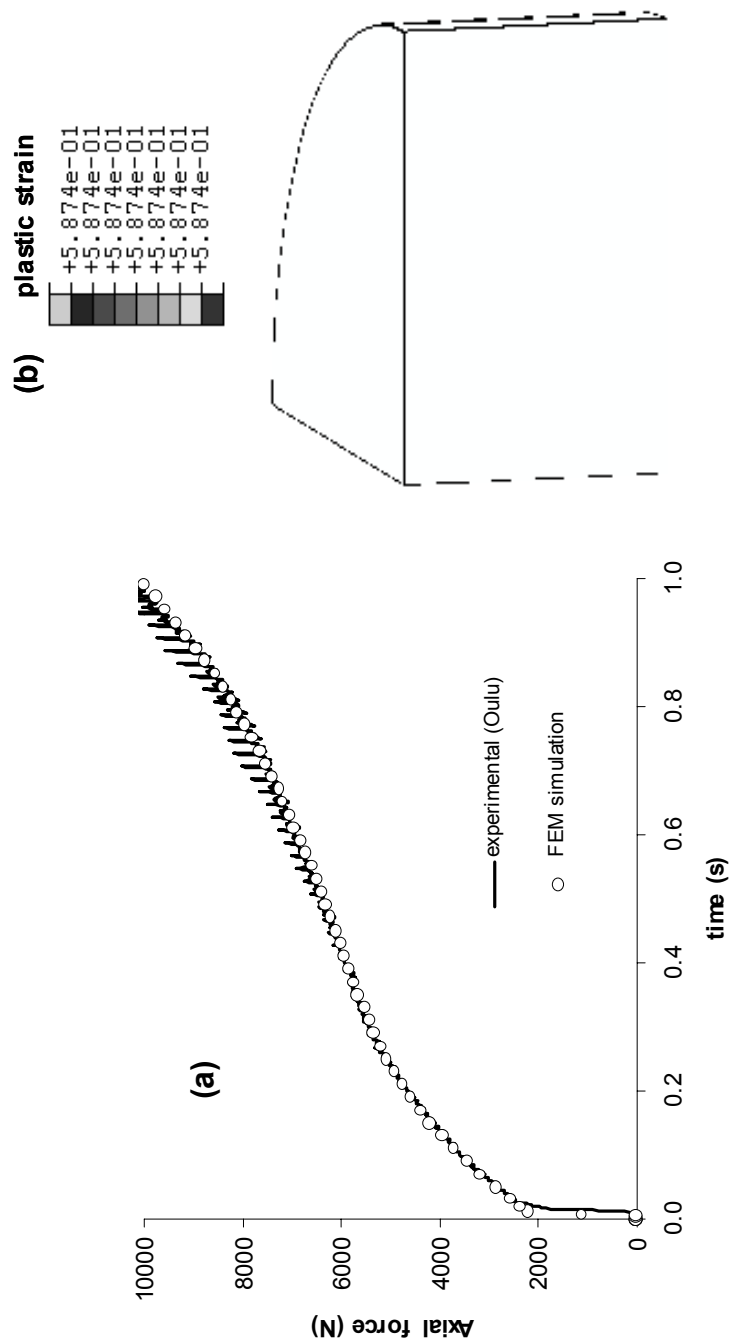


Fig. A3.4: FEM simulation of a Gleeble monotonic axisymmetric compression test, 55Cr3 medium-C steel at 1050 °C and 1 s⁻¹ : (a) comparison of experimental and simulation raw force results (b) detail of the octave section used in the FEM simulation at (constant) plastic strain 0.587.

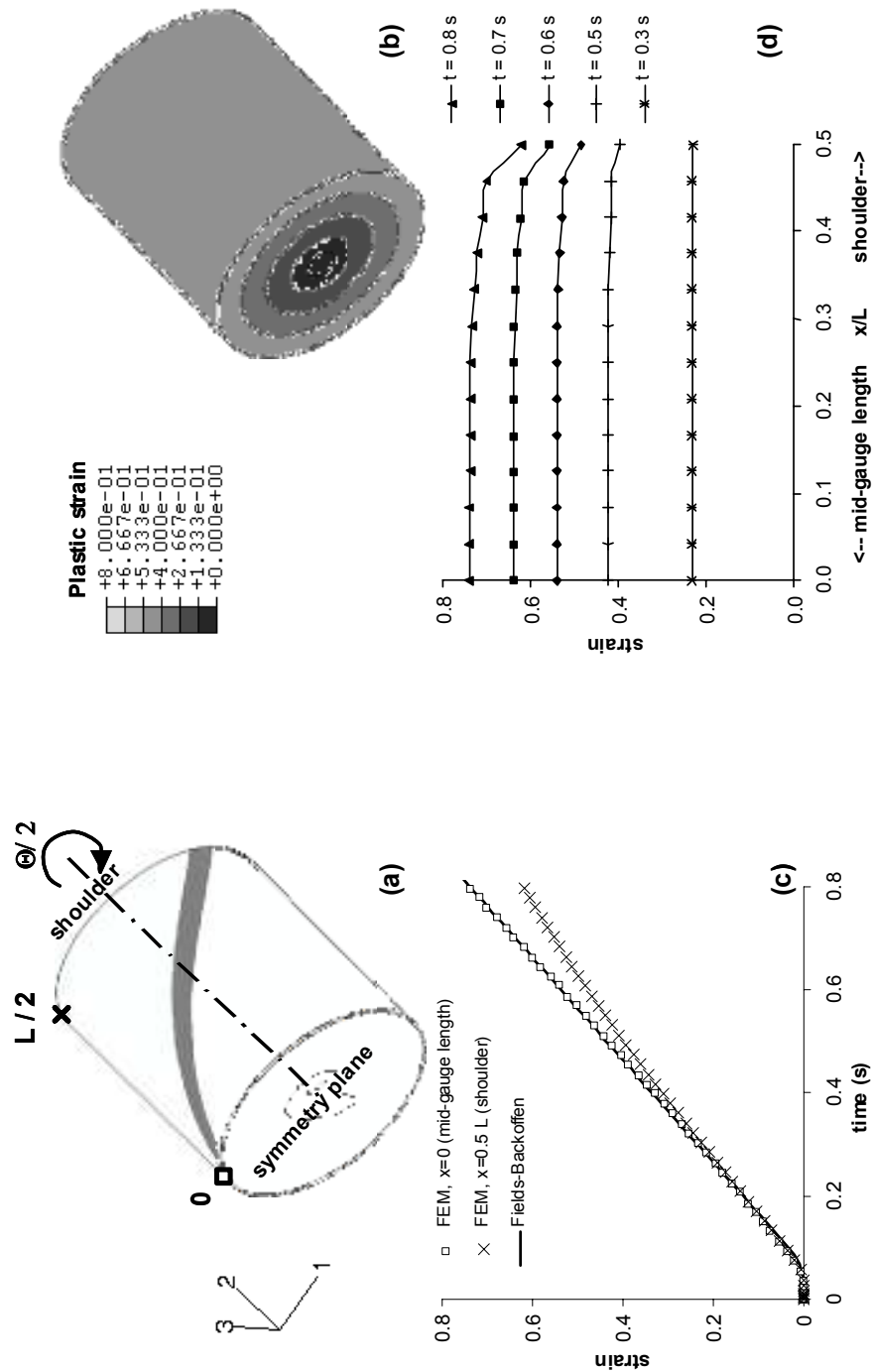


Fig. A3.5: FEM simulation of a monotonic torsion test, 55Cr3 medium-C steel at 1050 °C and 1 s⁻¹; (a) details of the half specimen and boundary conditions; (b) resulting local strain distribution; (c) comparison of local strain distribution and local FEM strains; (d) progress of strain localisation in time.

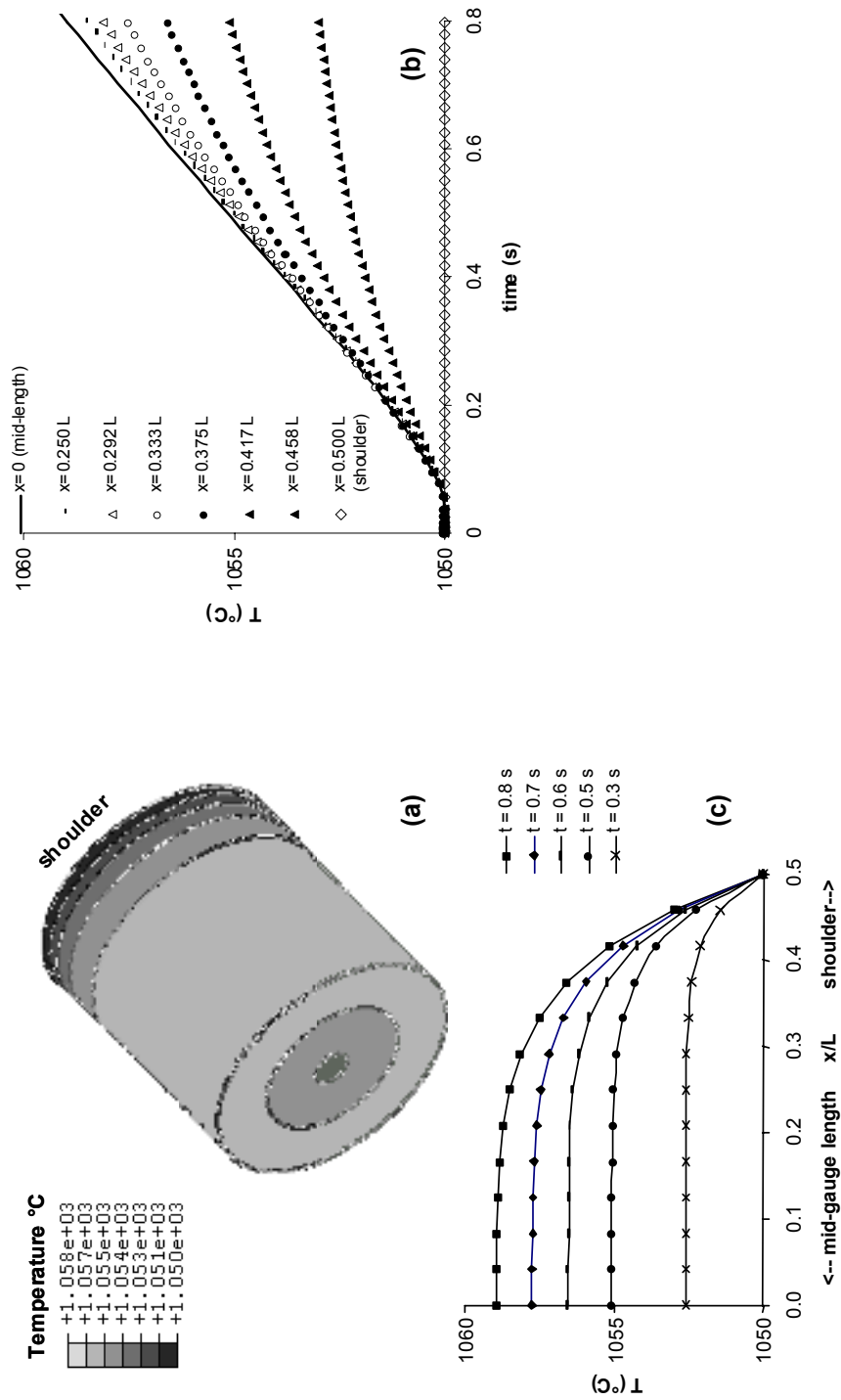


Fig. A3.6: FEM simulation of a monotonic torsion test, 55Cr3 medium-C steel at 1050 °C and 1 s⁻¹; (a) temperature distribution; (b) and (c): progress of thermal localization.

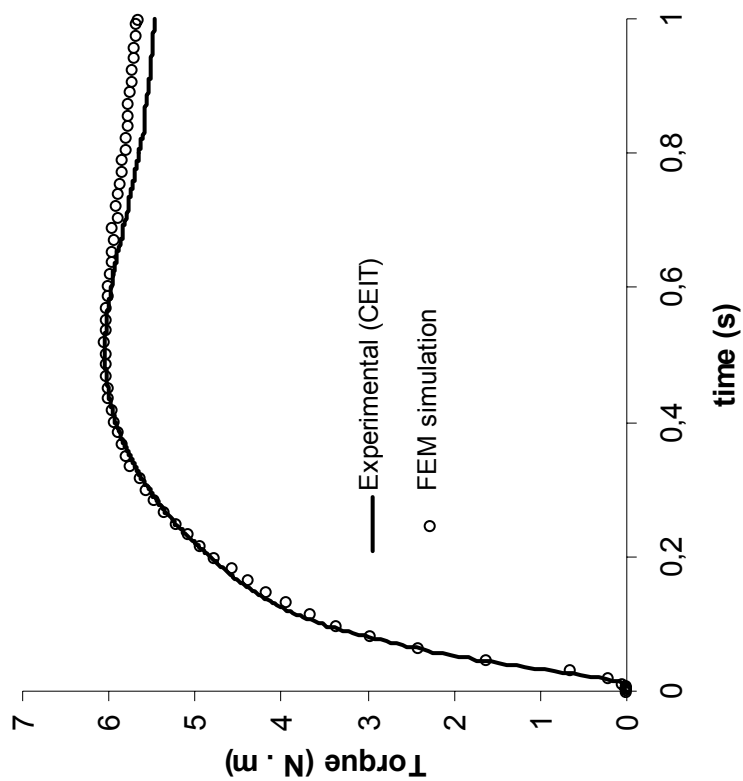


Fig. A3.7: FEM simulation of a monotonic torsion test, 55Cr3 medium-C steel at 1050 °C and 1 s⁻¹. Comparison of experimental and simulation raw torque results.

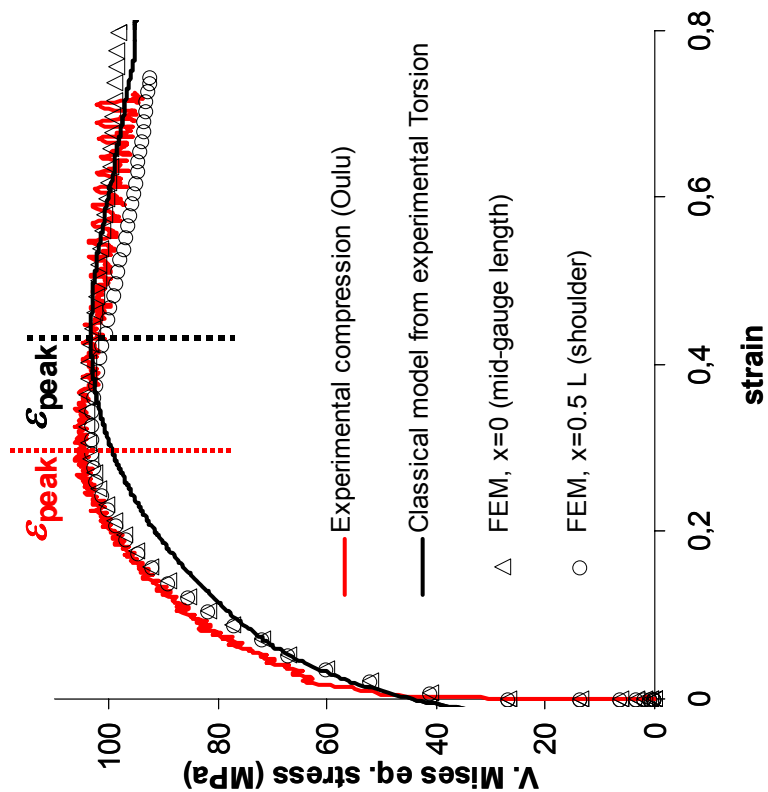


Fig. A3.8: Comparison of stresses from different techniques, monotonic torsion test, 55Cr3 medium-C steel at 1050 °C and 1 s⁻¹. Note the agreement between compression experiment length and the difference in strain to peak stress between compression and classical torsion analysis.

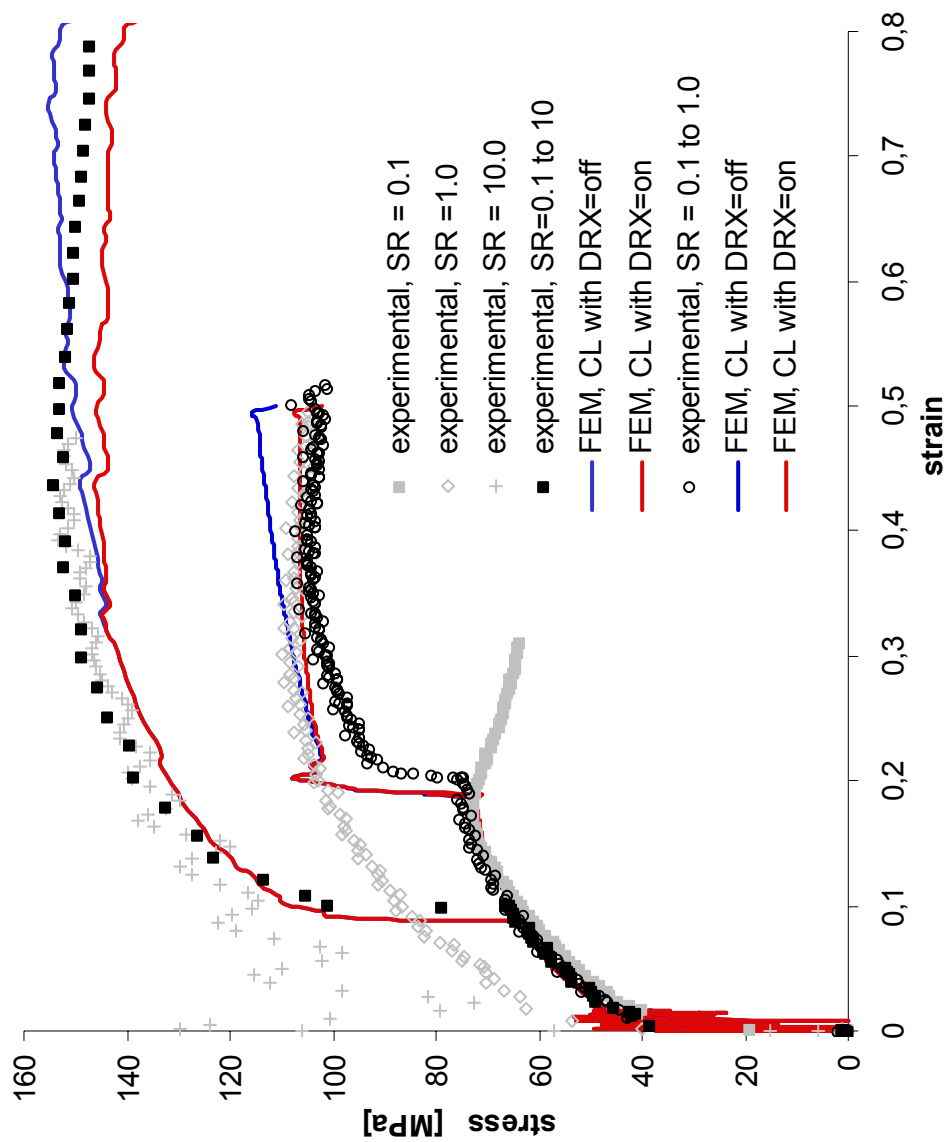


FIG. A3.9: Stepped increasing strain rate Gleeble compression, 55Cr3 medium-C steel at 1050°C. Comparison of the University of Oulu experiments and CSM-CL predictions. Note the effect of DRX.

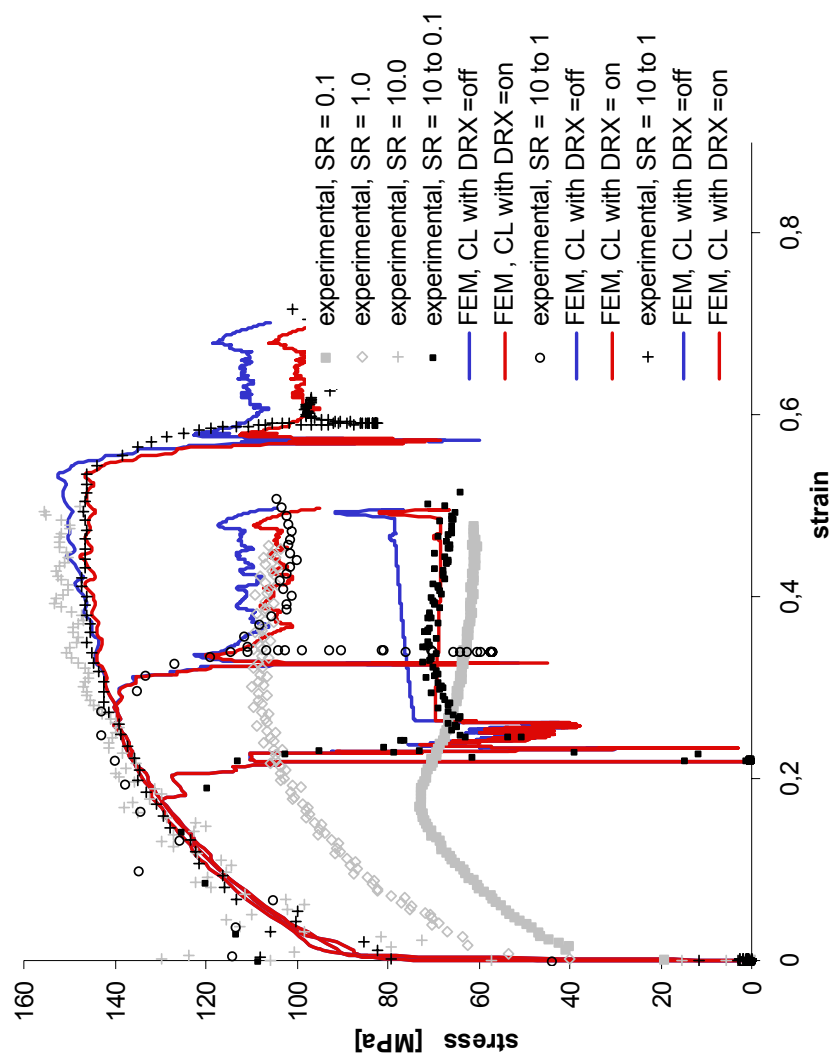


Fig. A3.10: Stepped decreasing strain rate Gleeble compression, 55Cr3 medium-C steel at 1050°C. Comparison of the University of Oulu experiments and CSM-CL predictions. Note the effect of DRX.

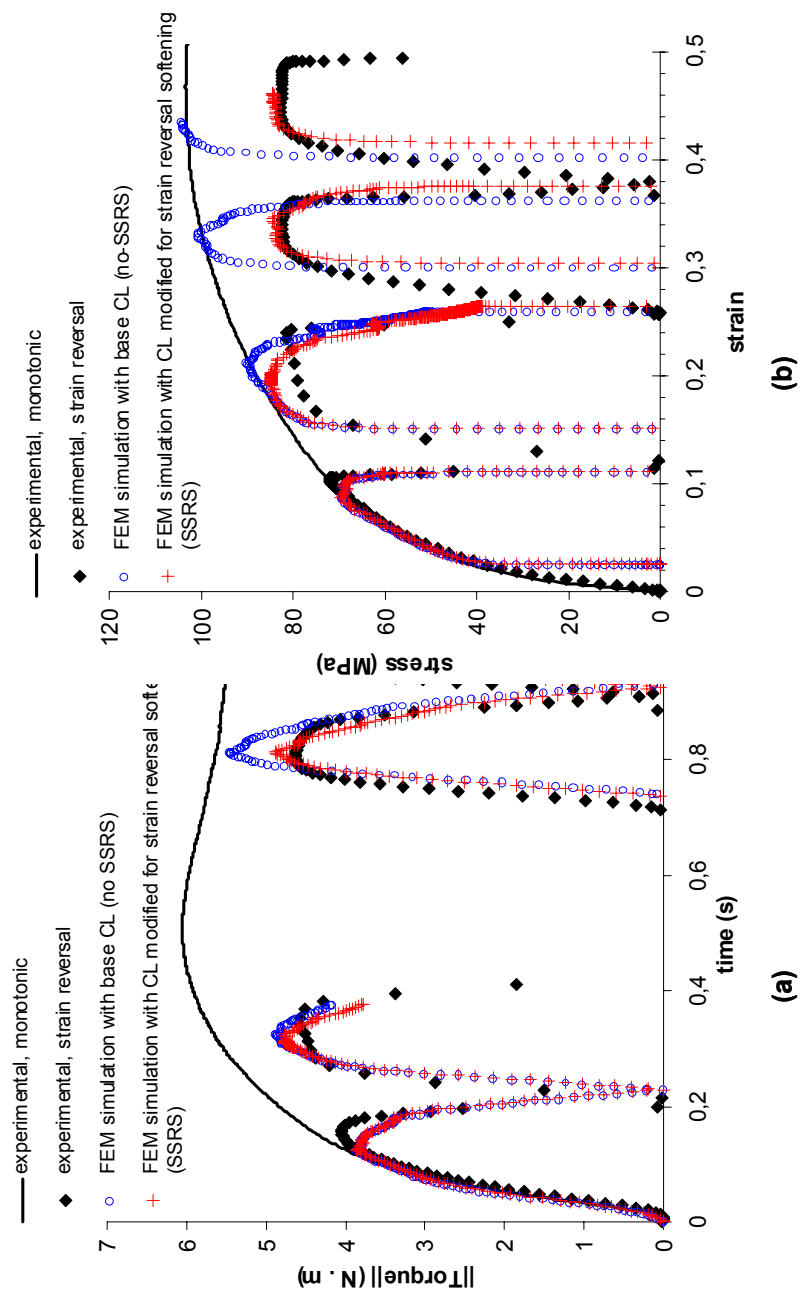


Fig. A3.11: FEM simulation of a torsion test, 55Cr3 medium-C steel at 1050°C and 1 s⁻¹. Comparison of monotonic and strain reversal results from CEIT experiments and CSM FEM simulations under the base (NO-SSRS) and modified (SSRS) constitutive laws:
 (a) absolute value of raw torque; (b) equivalent stress.

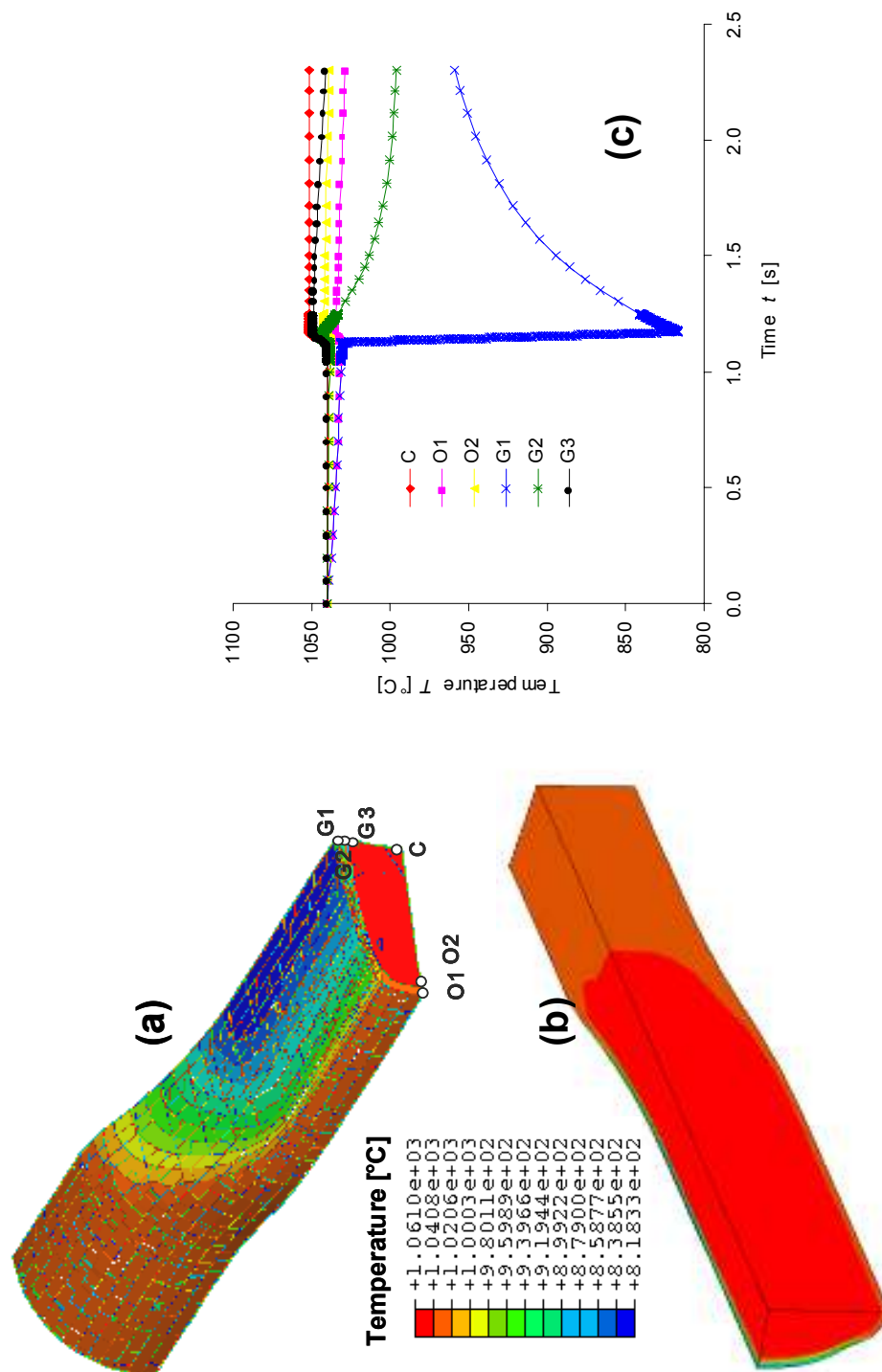


Fig. A3.12: Typical thermal distribution predicted by the FEM simulation, round-oval 8th pass of ORI-MARTIN bar mill.

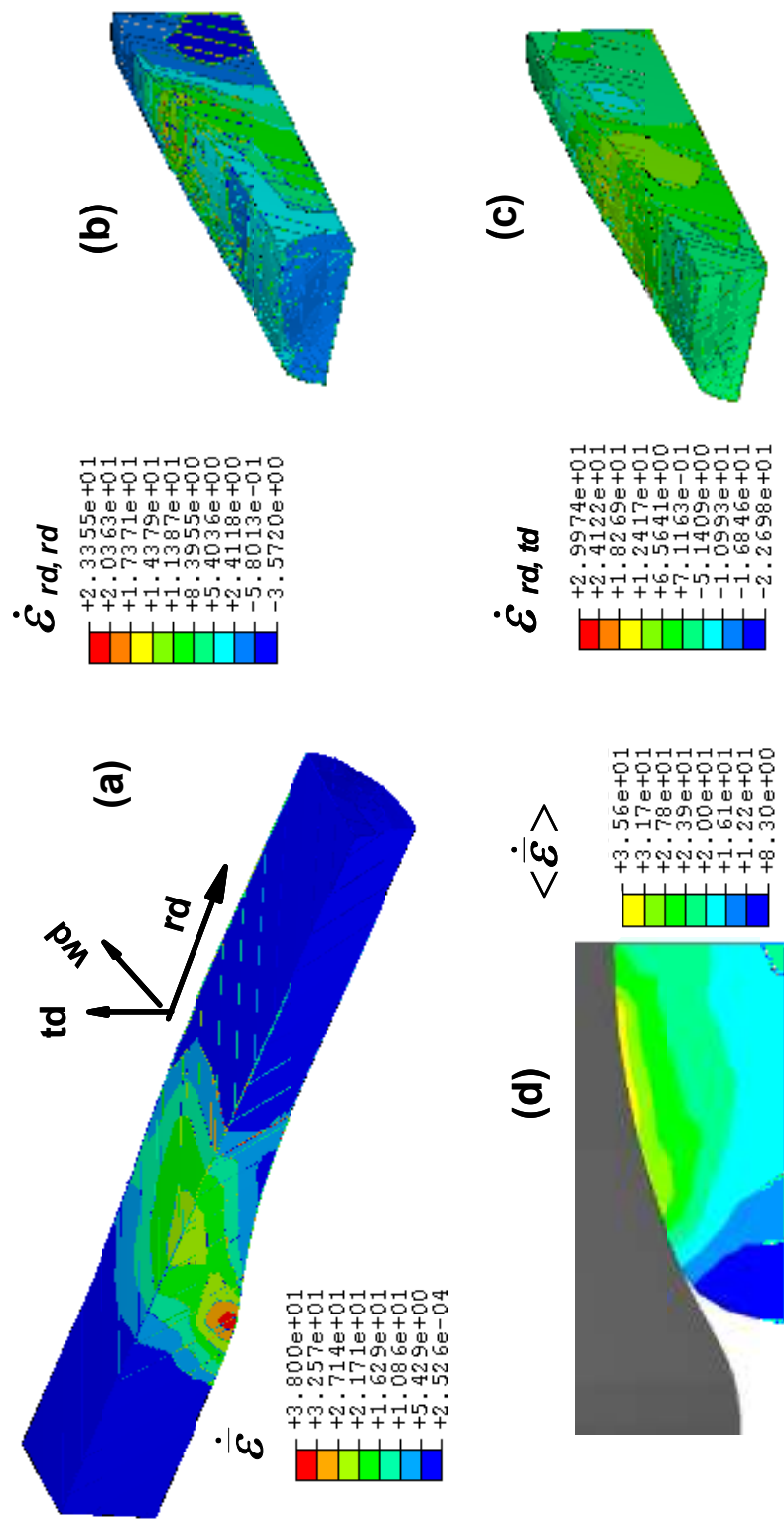


Fig. A3.13: Typical strain rate distribution predicted by the FEM simulation, round-oval 8th pass of ORI-MARTIN bar mill: (a) total effective strain rate; (b) strain rate normal component due to the elongation in the rolling direction; (c) shear strain rate component in the rolling-thickness direction; (d) average strain rate; rd =rolling direction, td =thickness direction, wd =width direction.

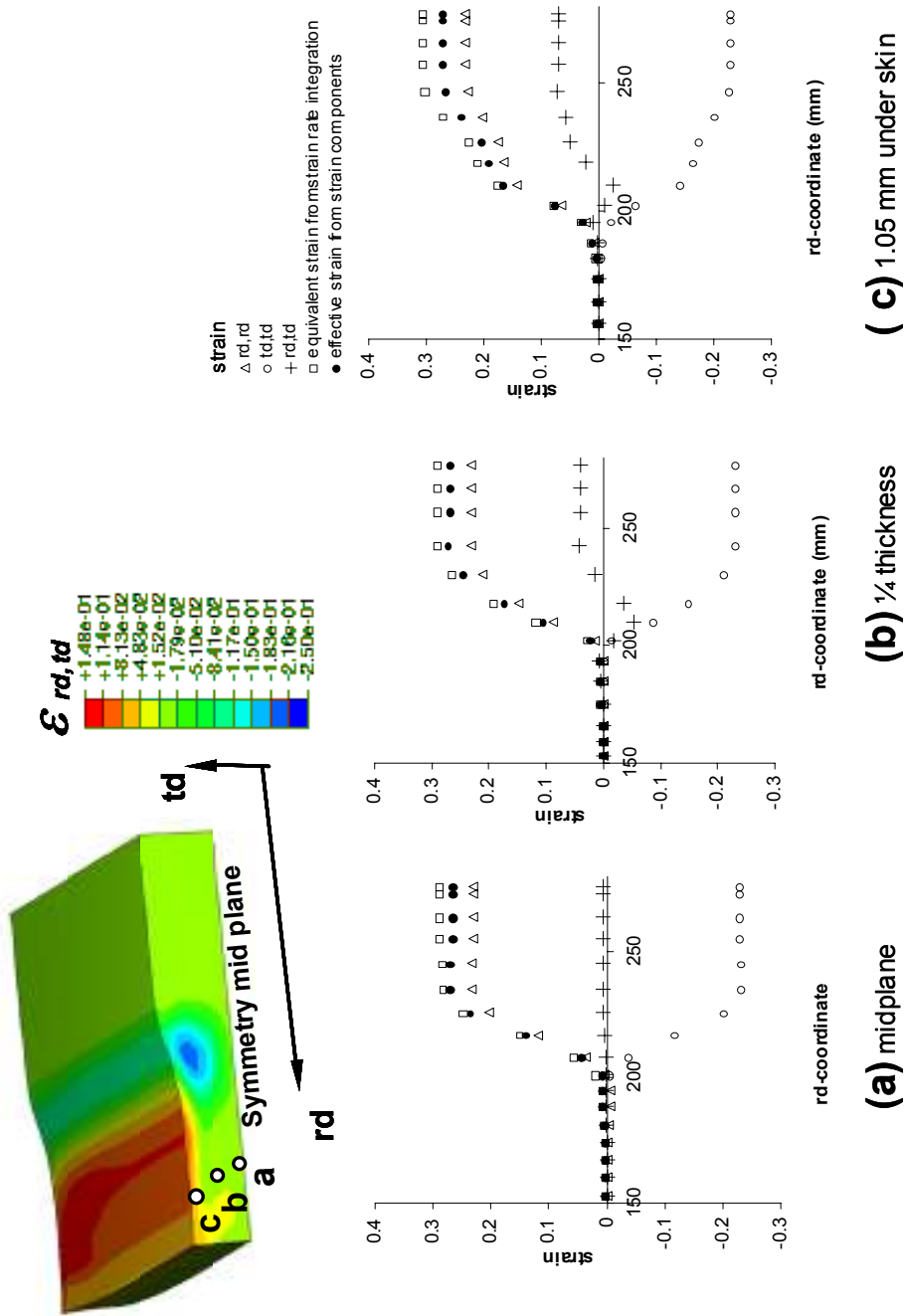


Fig. A3.14: Evolution of strain components and their equivalent measures within the roll gap. Flat rolling of duplex stainless steel 2304. Entry thickness= 20 mm, rolling reduction=25.8%, friction coefficient 0.35; rd =rolling direction, td =thickness direction, wd =width direction.

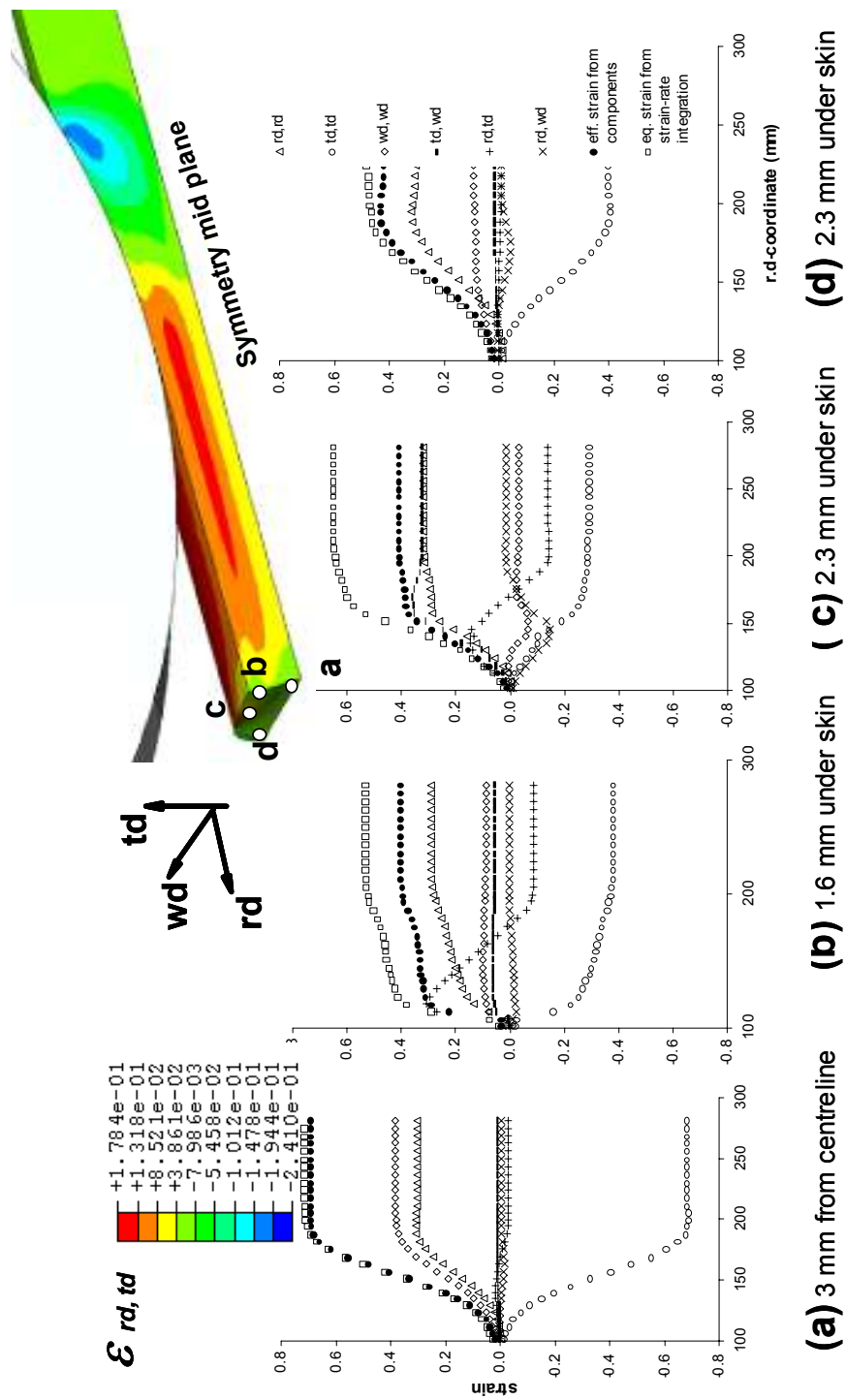


Fig. A3.15: Evolution of strain components and their equivalent measures within the roll gap. Round-oval 8th pass, medium-C 55Cr3 steel. Entry diameter 55 mm, rolling area reduction=38%, friction coefficient 0.35; rd =rolling direction, td =thickness direction, wd =width direction.

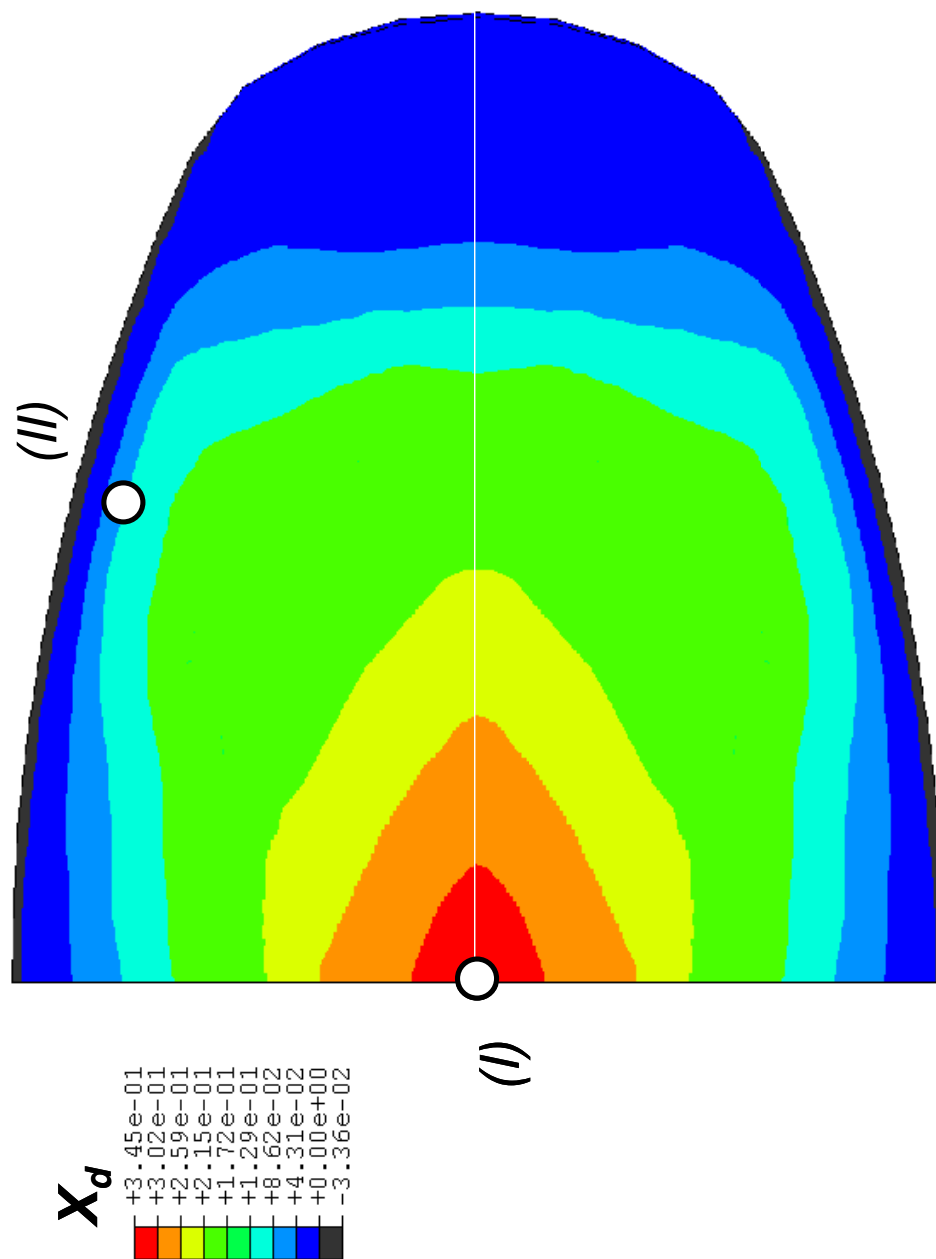


Fig. A3.16: Recrystallised volume fraction predicted by the model at the exit of the 8th round-oval pass, $T=1040^{\circ}\text{C}$, rolling speed ≈ 1.73 m/s; (I) and (II) are streamlines at the roll gap exit; medium-C 55Cr3 steel.

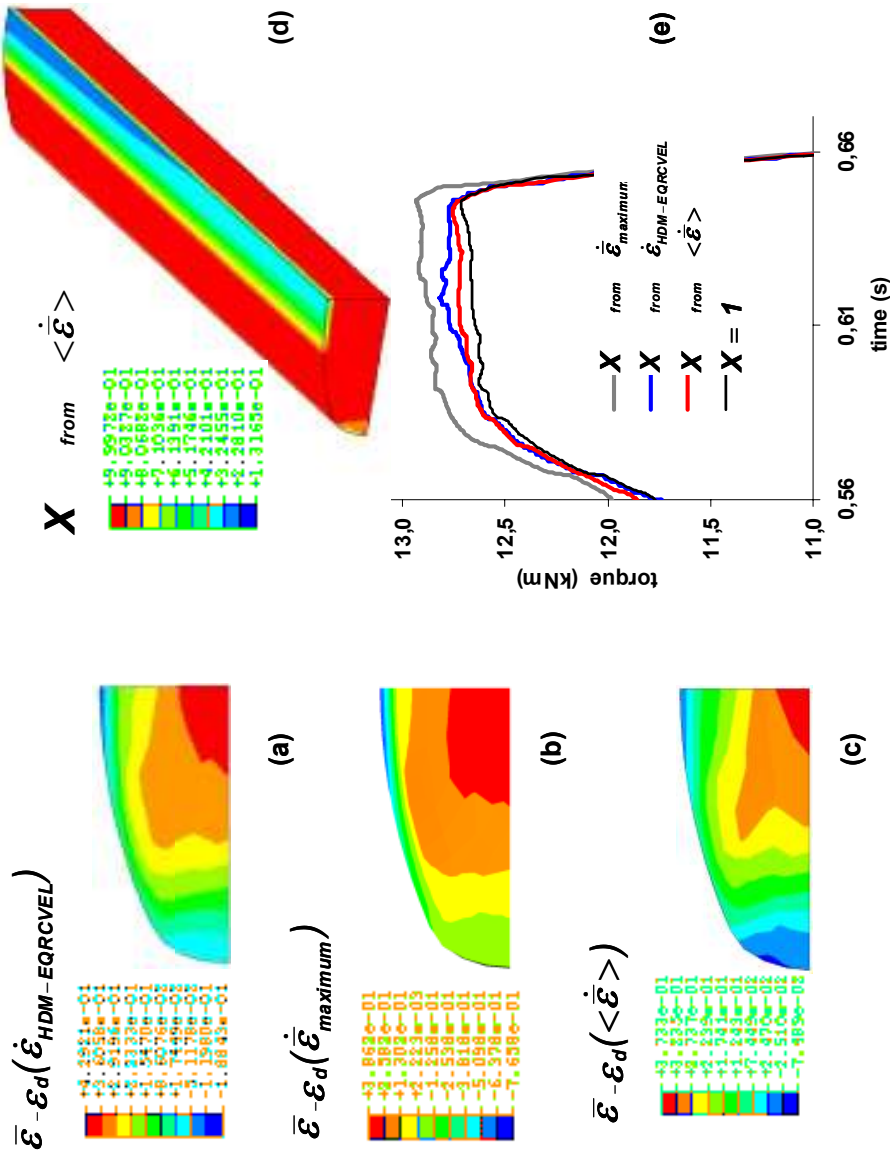


Fig. A3.17: FEM softening predictions for the 8th-9th interstand of ORI-MARTIN bar mill: (a)-(c) effect of different prior strain rate measures on the difference between strain and DRX strain; (d) evolution of recrystallisation in the interpass; (e) effect of the prior strain rate choice on the resulting torque; medium-C 55Cr3 steel.

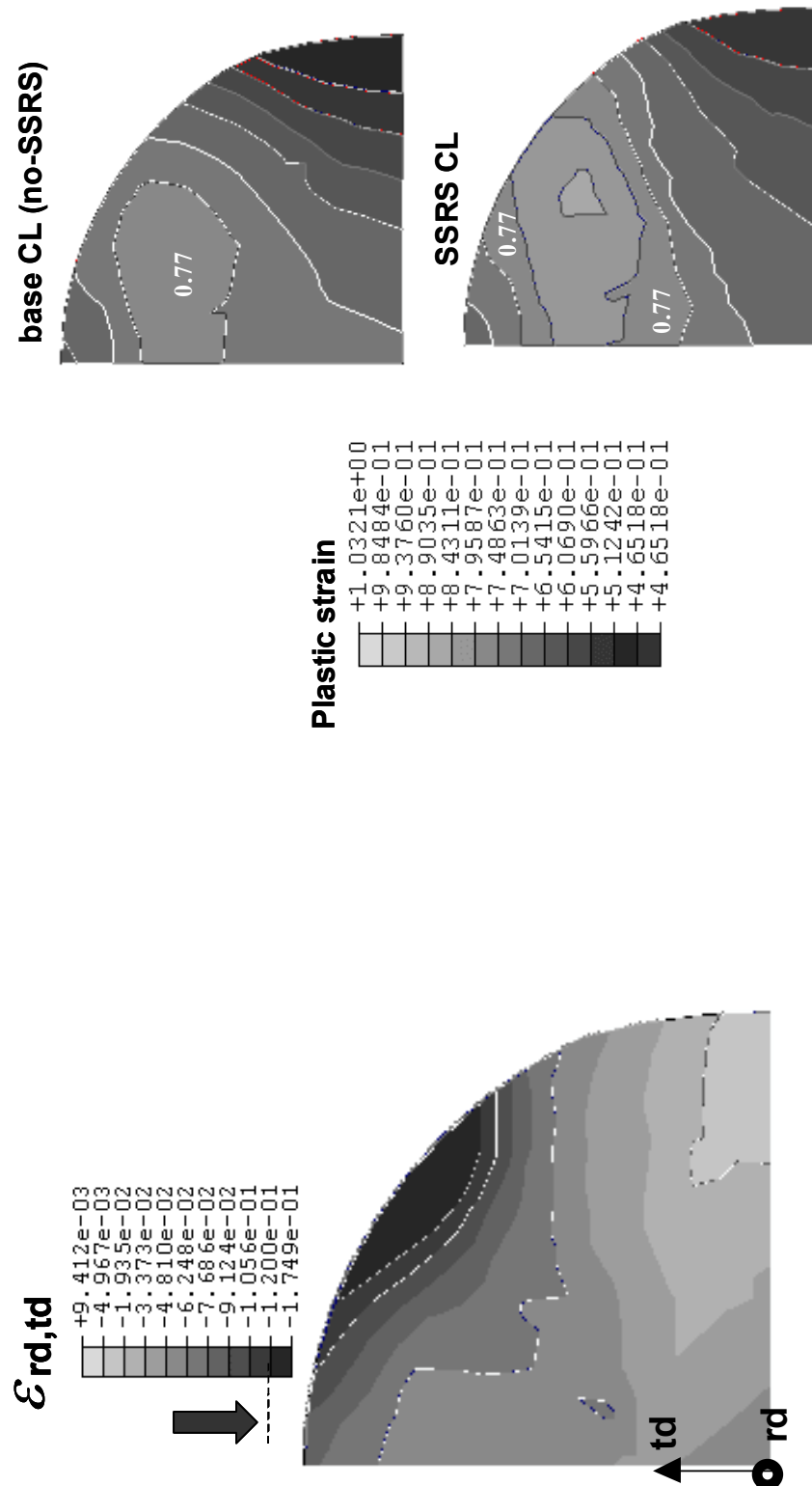


Fig. A3.18: Area affected by shear strain reversal during 7th stand oval-round pass, area reduction 27%, initial section 39 x 90 mm.

Fig. A3.19: FEM prediction of cross-section local strains for the conventional (NO-SSRS) and modified (SSRS) Constitutive Law; 7th oval-round pass; medium-C 55Cr3 steel.

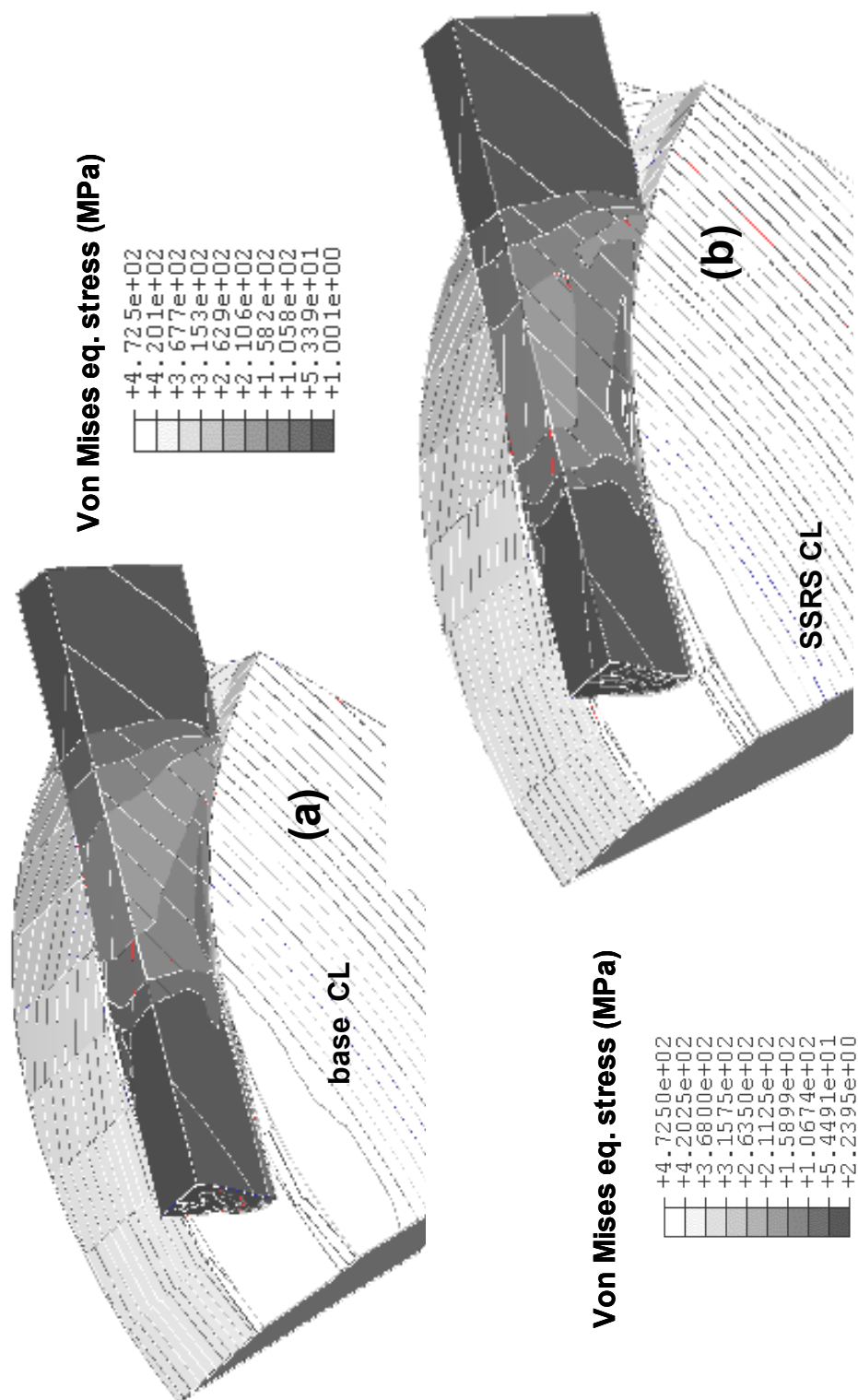


Fig. A3.20: FEM prediction of local stress during rolling for the conventional (NO-SSRS) and modified (SSRS) Constitutive Law; 7th oval-round pass; medium-C 55Cr3 steel.

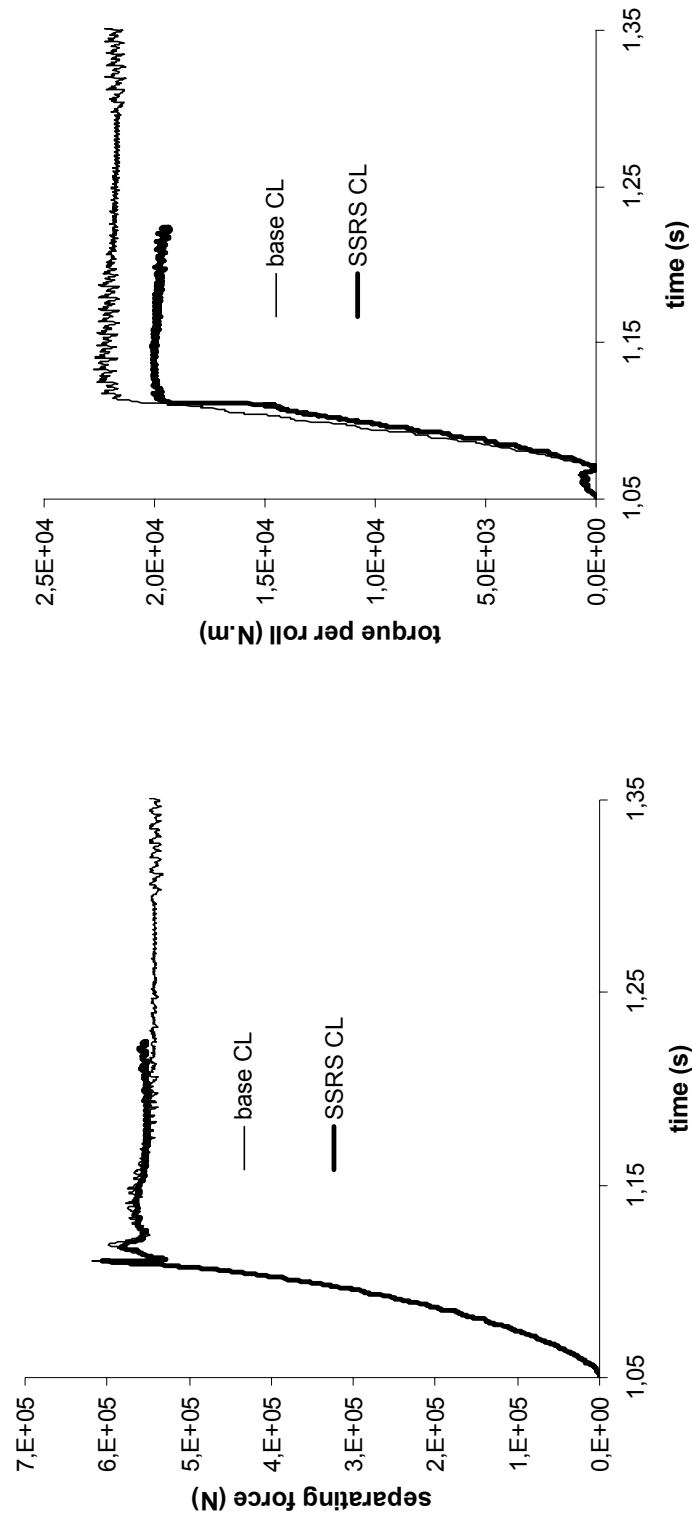


Fig. A3.21: Comparison of the Rolling Force and Rolling Torque predicted by the conventional (NO-SSRS) and modified (SSRS) Constitutive Law for the 7th oval-round pass; medium-C 55Cr3 steel.

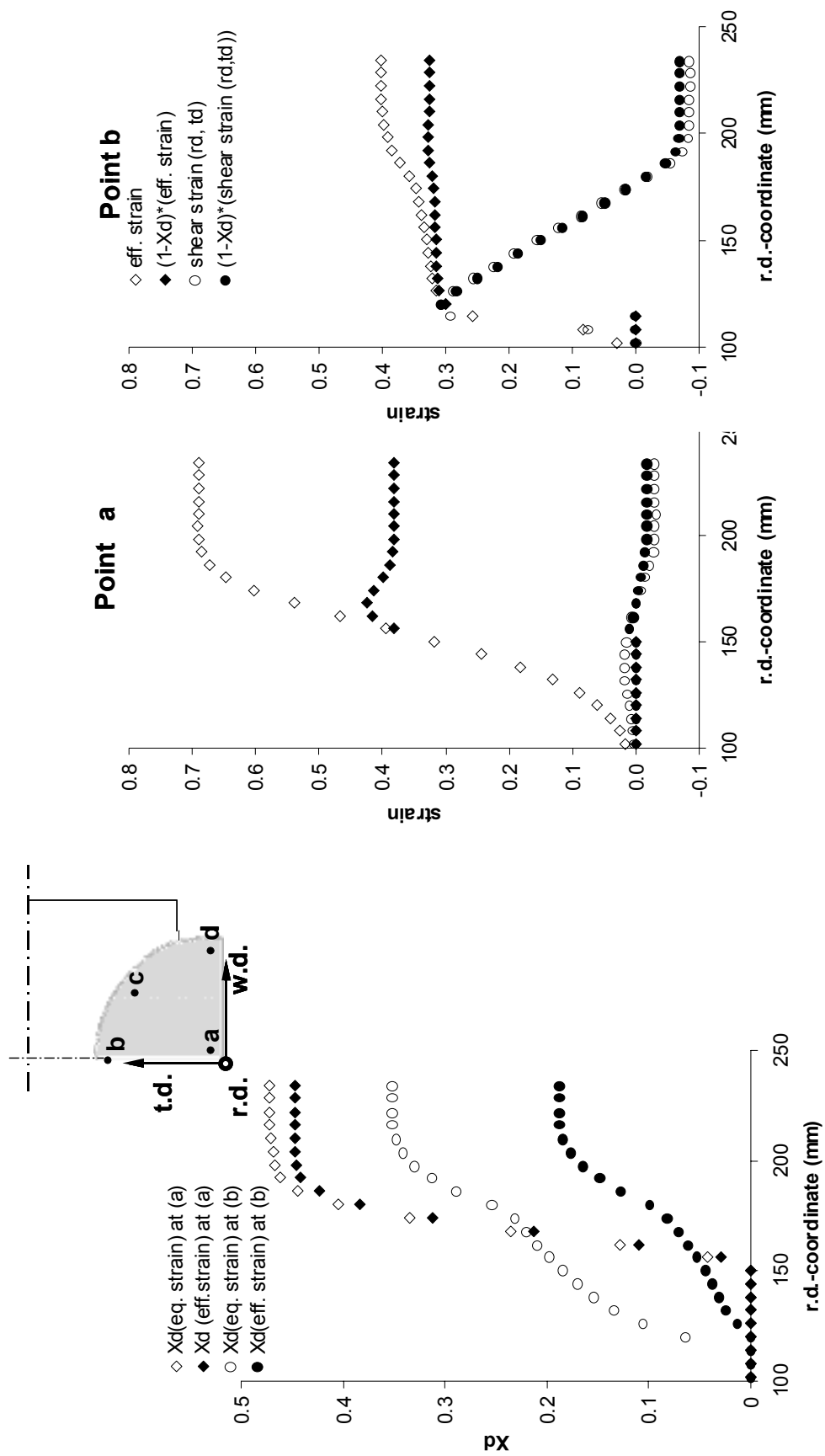


Fig. A3.22: Effect of shear strain reversal on DRX; 7th oval-round pass, medium-C 55Cr3 steel.

Fig. A3.23: Effect of DRX on shear strain reversal; 7th oval-round pass, medium-C 55Cr3 steel.

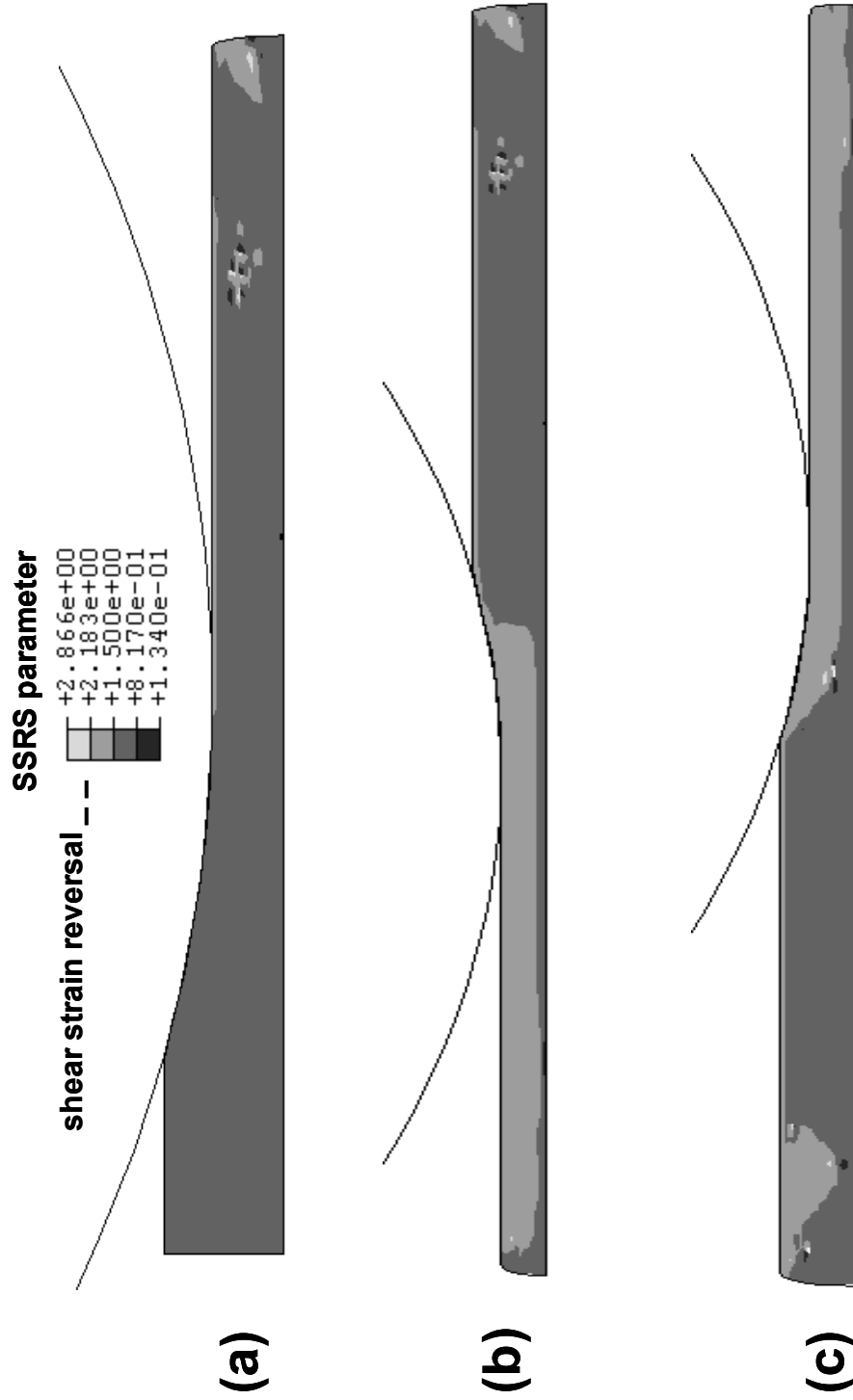


Fig. A3.24: Evolution and through-passes propagation of the area affected by the strain reversal in a non-recrystallising 2304 steel during flat rolling: (a) first pass; (b) second pass in reversing mode; (c) second pass in tandem mode.

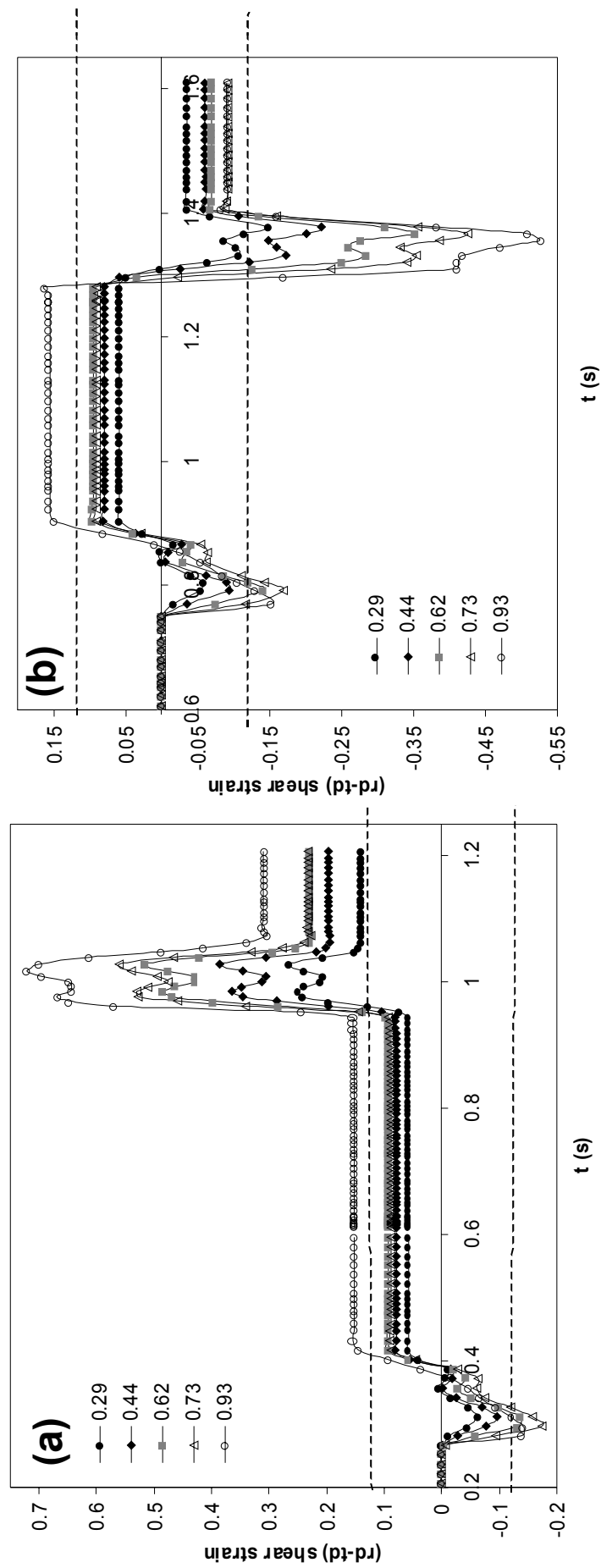


Fig. A3.25: Evolution of the shear strain component during two consecutive passes at several through thickness:
 (a) pass 1— reversing pass 2; (b) pass 1— tandem pass 2. Flat rolling, 1000°C, two 40% reduction
 passes from $H_0 = 35$ mm.

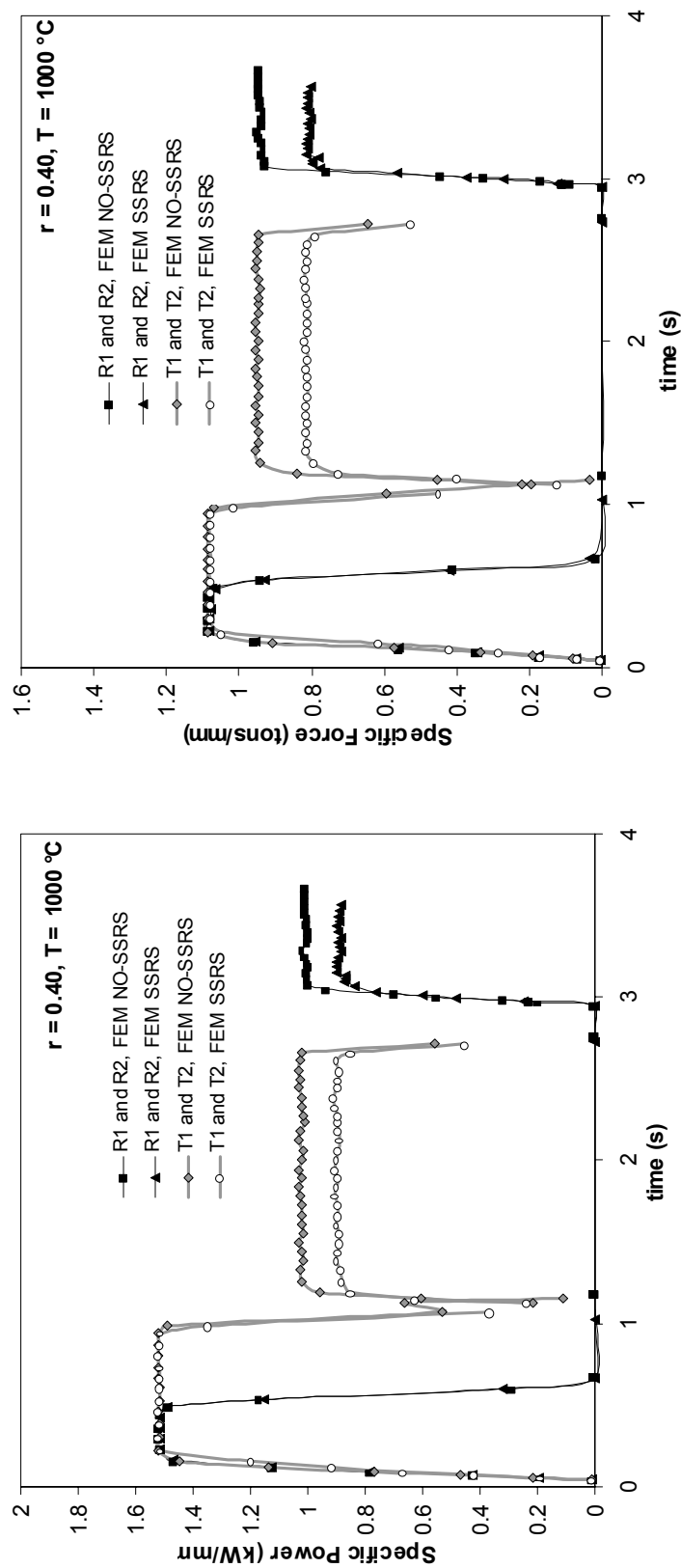


Fig. A3.26: Comparisons of the predicted power consumption and specific roll force for two consecutive roughing and tandem passes under the NO-SSRS and SSRS constitutive hypotheses in the second pass. Non-recrystallising 2304 steel, 1000°C, two 40 % reduction passes from $H_0 = 35$ mm.

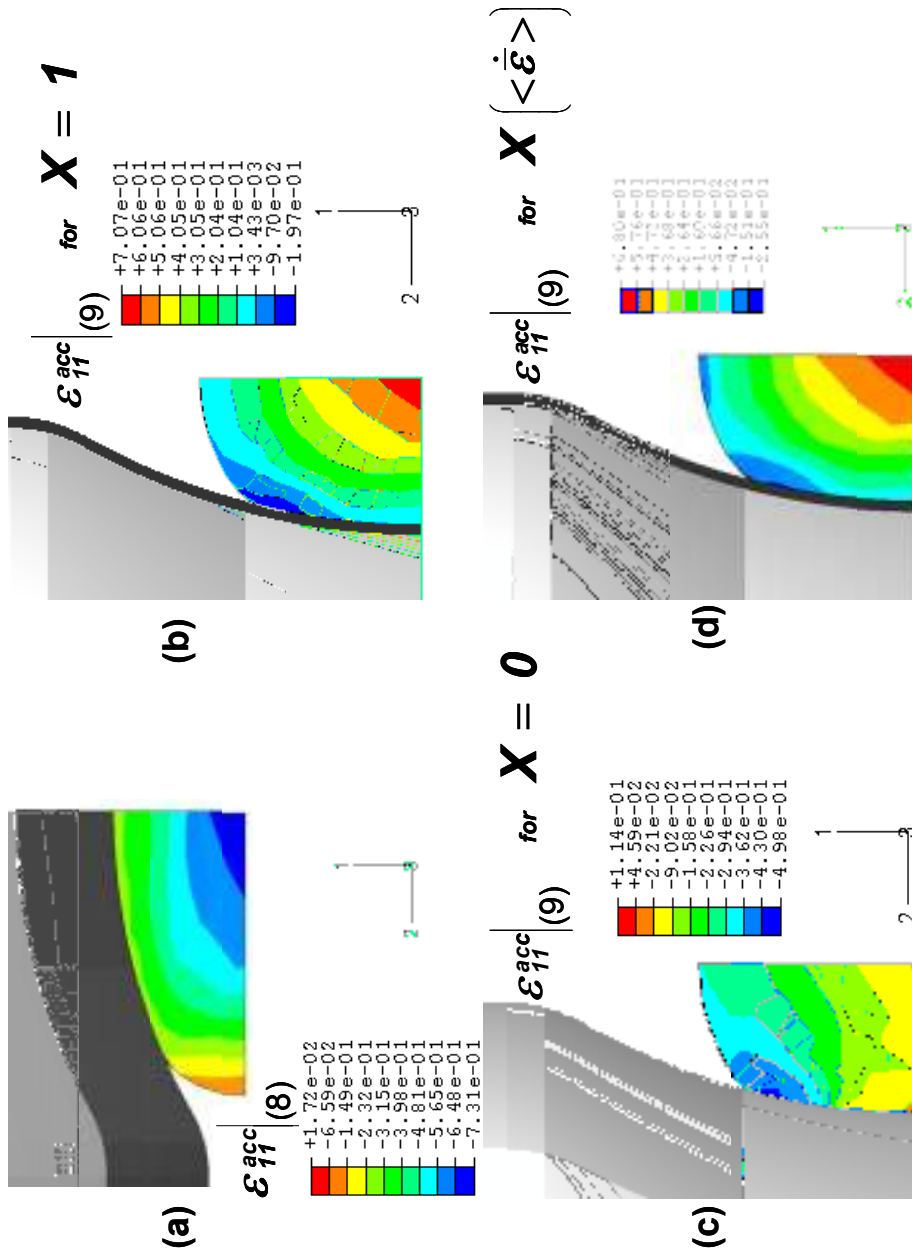


Fig. A3.27: Accumulation of strain components between consecutive H-V stands. Case study: normal strain component ε_{11}^{acc} in the thickness direction; (a) round-oval 8th pass; (b) round-oval 9th pass after full interpass recrystallisation; (c) oval-oval 9th pass without recrystallisation (hypothetical case not applicable to the medium-C steel under study); (d) oval-oval 9th pass after recrystallisation in the interstand computed with the medium prior strain rate.

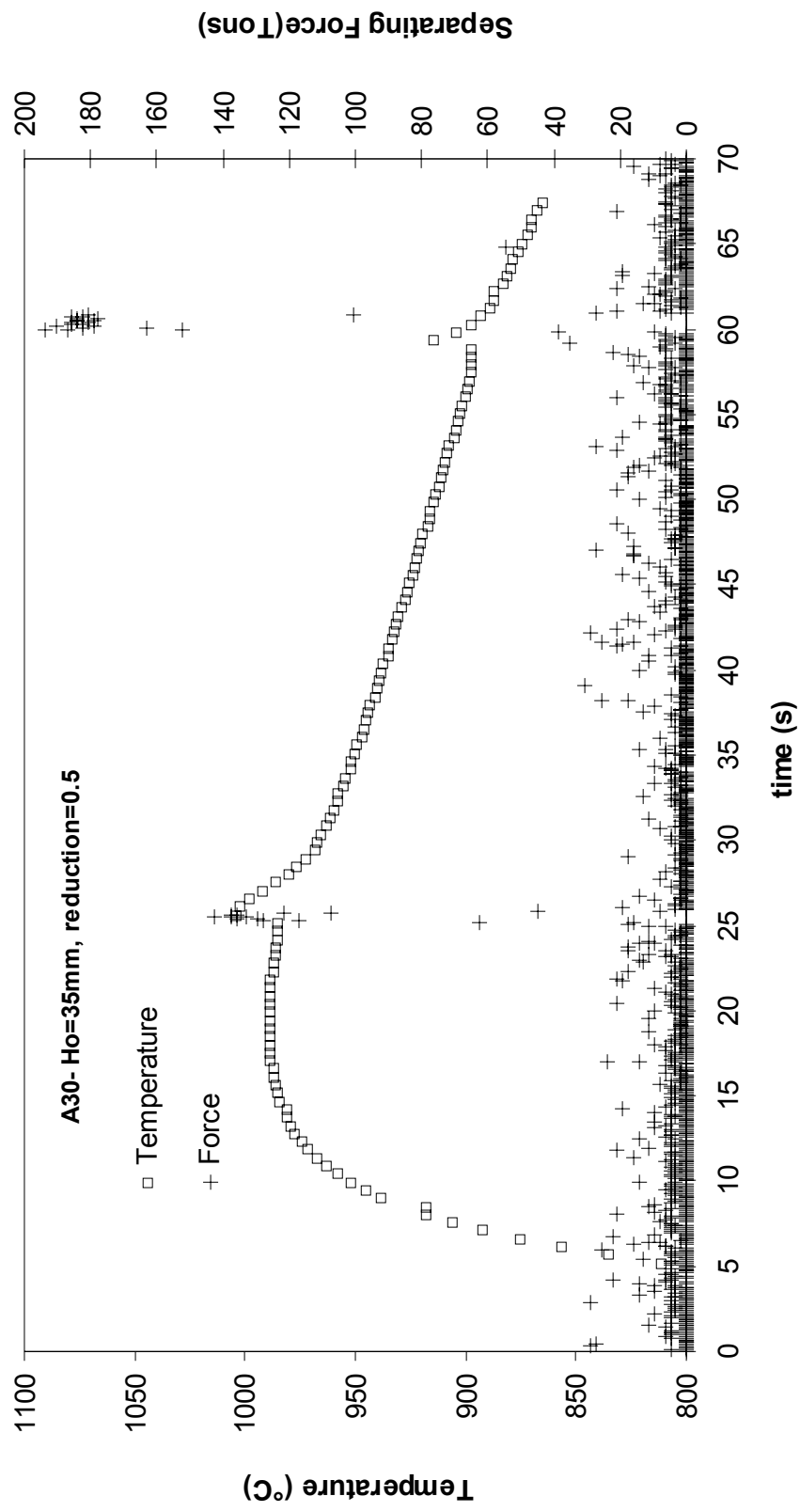


Fig. A3.28: Experimental temperature evolution during the two rolling passes in case study A30 (Table A3.9)

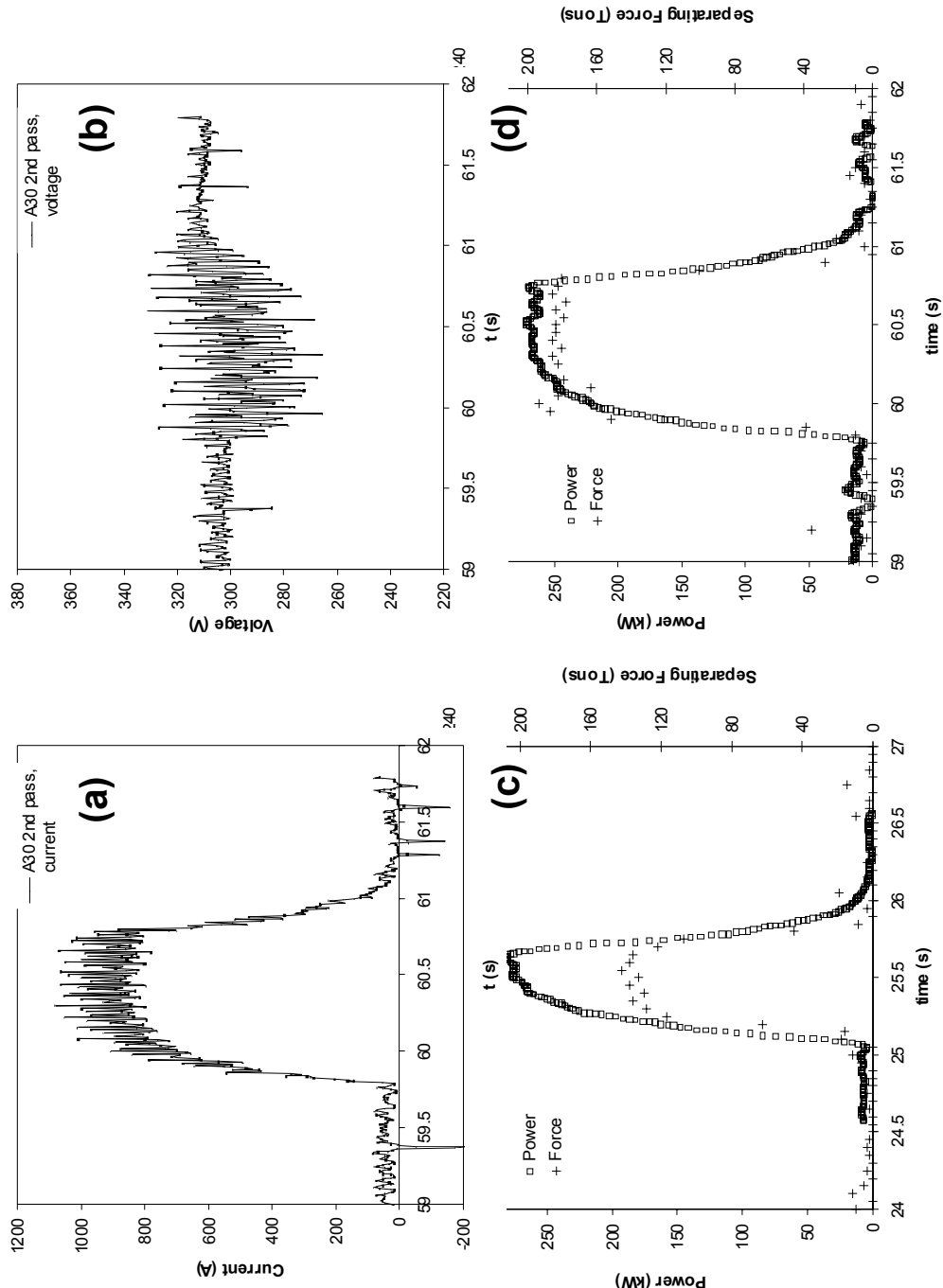


Fig. A3.29: Typical experimental rolling forces and power consumption data: (a) and (b) raw electrical data; (c) and (d) load cells data for forces and FT-processed power data for two consecutive passes (case study A30).

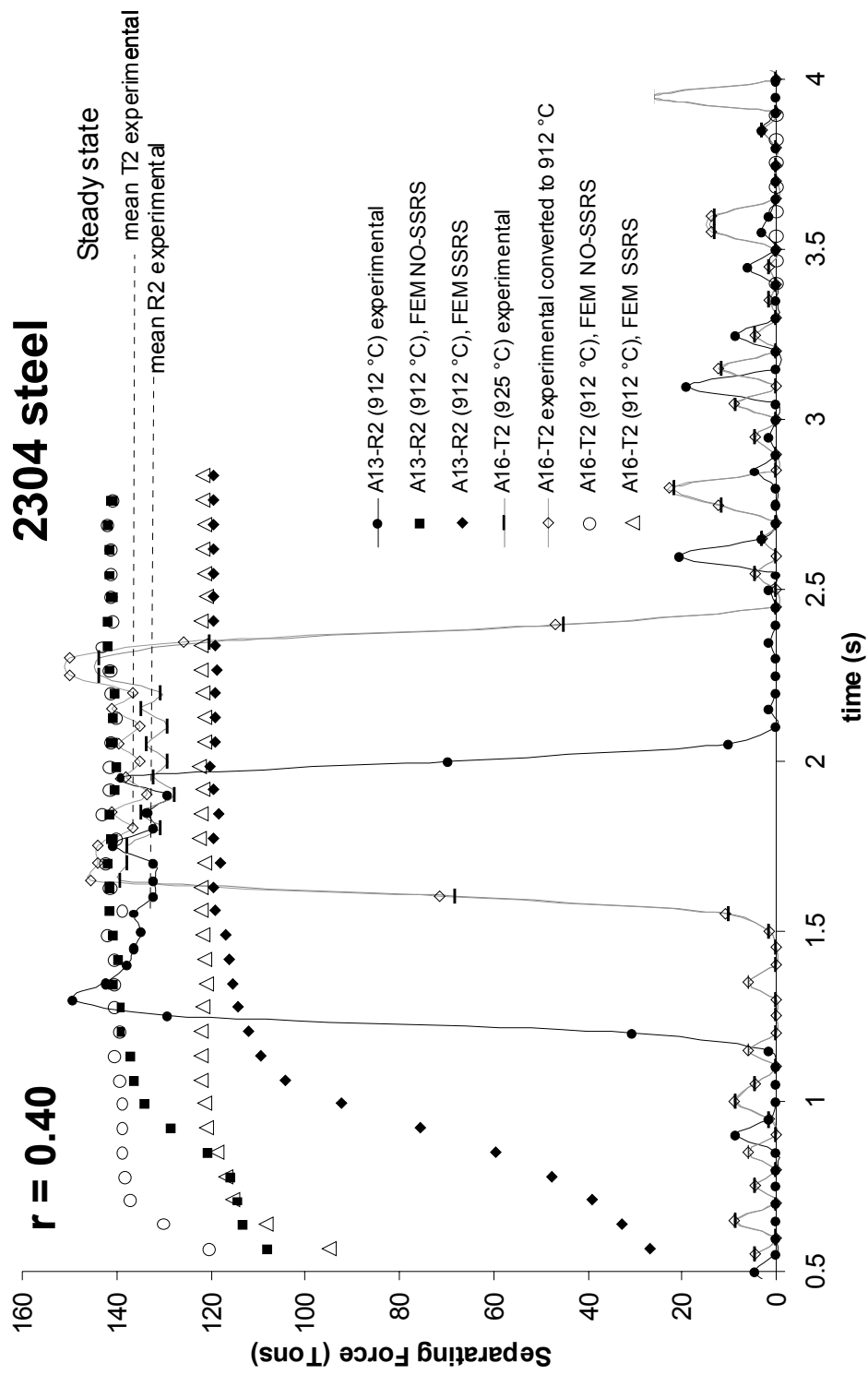


Fig. A3.30: Comparison of experimental and FEM predicted rolling force for the second pass. Flat rolling, 1000°C, pass reduction 40% from $H1=21$ mm.

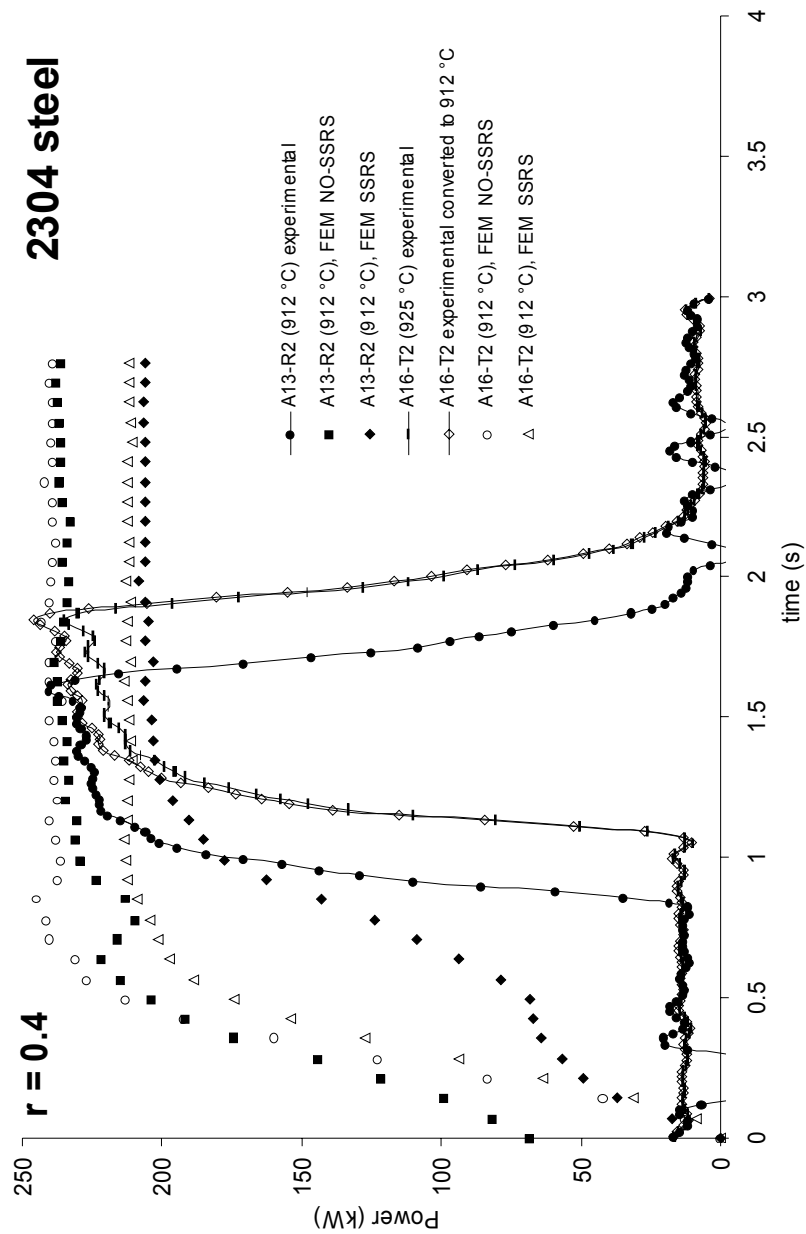


Fig. A3.31: Comparison of experimental and FEM predicted power consumption for the second pass. Flat rolling, 1000°C, pass reduction 40% from $H_1=21$ mm.

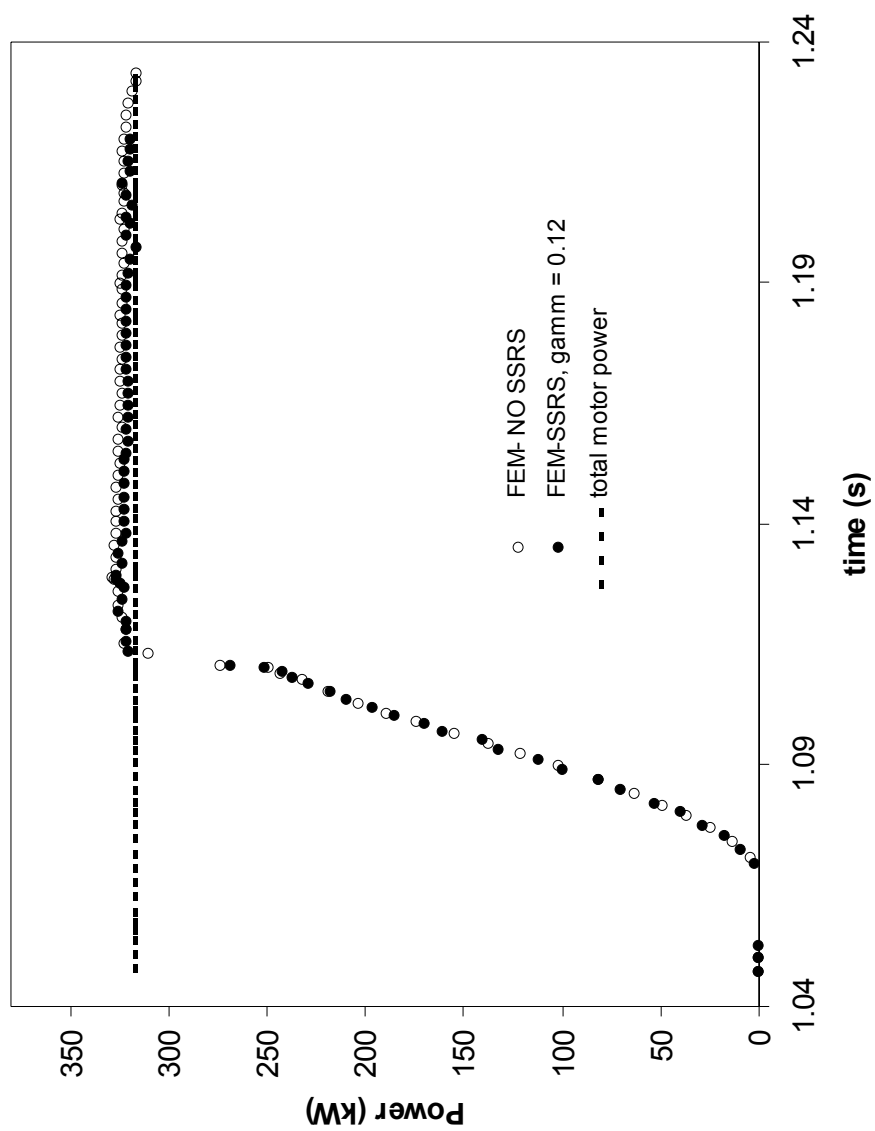


Fig. A3.32: Comparison of FEM prediction and experimental measurement, rod rolling of medium-C steel 55Cr3, Temperature = 1020 °C, roll diameter = 0.433 m, Area Reduction = 23%. Conversion efficiency factor = 0.85

APPENDIX 4

Detailed Technical Report by MEFOS Author: M Karlberg

A4.1 Material of interest

In this project two different materials has been considered, C-Mn-Nb and Cr steels. Where the most relevant measurements are yield stresses versus temperatures and strain rates in the range of interests. One should though point out that if microstructure modelling is of interest, data for these models as well could be important but these data has not been used in this project. However, the first stage in this project was to use the data for modelling of the yield stresses based on dislocation and vacancy theories which means that a lot of work can be done by deformation data since magnitudes of the microstructure parameters generally are known and does not differ so much between different materials. Yield stress data for C-Mn-Nb steel has been collected at the pilot plant in Freiberg where it has been measured for a number of different strain rates at different temperatures. The collected data is especially important for the optimisation of parameters in the existing constitutive model. When parameters have been optimised, validation of the constitutive model for the C-Mn-Nb steel can be performed for compression tests as well as for the benchmark example.

A4.1.1 C-Mn-Nb

Data has been processed at the partner TU Freibergs mill facility in Germany where samples are measured at a several temperatures between 600°C up to 1200°C but only temperatures of interests are 800, 850, 900, 1000, 1100 and 1200°C. Strain rates were sampled at 0.1, 1 and 10/s. This kind of data is corresponding to quite low strain rates which typically for the benchmark case is around 20/s. Anyhow the strongest dependence of the strain rate on the yield stress is usually in the lower strain rate regions, lets say between 0.1 up to 10/s whereas this data still gives a lot of valuable information when optimising the parameters in the model.

A4.1.2 Cr-Mo-V steel

Material of interest in the Nordic committee at MEFOS and is categorized as tooling steel. Where the characteristics are high toughness against cyclic temperature changes, high strength at elevated temperatures, easy to harden and machine tooling. Typical yield strengths for this material are about 1400 MPa at room temperatures.

Measurements has been sampled at the Swedish institute of material research (SIMR) by compression tests at the strain rates 0.8, 3.5, 10 and 30/s and at temperatures from 800°C up to 1200°C, at every even hundred degrees. All samples have been exposed to a prior austenitisation for 10 hours at 1250°C. This data originally was produced for a similar project at SIMR and therefore was very convenient to use for almost the same objectives in this project.

A4.2 Constitutive models

Some of the models presented in this report are the same that were used in a former project, Ref. [A4.1]. The models used in that project were mainly basic equations for description of the evolution of the dislocation density and the vacancy concentration, recovery processes was the same as in this project further explained in Section 2.1. In order to develop and

further increase the complexity of the models a cooperation between SIMR, MIKRAB and MEPOS has been put together. The reason for this was that SIMR together with MIKRAB has great experience and knowledge in microstructure modelling and material science.

A4.2.1 Dislocation density and vacancy formulations

Dislocation density evolution is generally described as the difference between generated mobile dislocations which acts like carrier of the plastic strain and immobilised dislocations. The generation of dislocations is represented by the following relationship:

$$\frac{d\rho_{gen}}{dt} = \frac{m \cdot \dot{\epsilon}}{b \cdot L} \quad \dots (A4.1)$$

Where m denotes the Taylor factor, a generalised correcting factor for the critical stress in a slip system correlated to the external applied stress, this parameter is slightly dependent of the strain as also the Burgers vector denoted b is, L is the mean free slip distance for mobile dislocations. Burgers vector usually is considered to be linearly dependent of the temperature by e.g. the following equation:

$$b = b_0 \cdot (1 + T \cdot C) \quad \dots (A4.2)$$

T is the temperature, C a slope constant and finally b_0 the initial burgers vector.

The mean free slip distance is described by Equation (A4.3) as follows.

$$\frac{1}{L} = \frac{1}{L_0 \cdot R_0} + \frac{\sqrt{\rho}}{k_L} \quad \dots (A4.3)$$

Where the slip distance L , is inverted proportional against the square root of the dislocation density, ρ . L_0 is the relative level of the slip distance, R_0 and k_L are proportionality constants needed to be optimised. The mean free slip distance can best be described as the mean distance where a dislocation can move without meeting any obstacles like other dislocations, precipitations or grain boundaries. The recovery process is dependent of several factors where dislocation climbing is one of the dominant ones especially when hot rolling is considered. Recovery of the dislocation density:

$$\frac{d\rho_{rec}}{dt} = M \cdot c \cdot \rho^2 \quad \dots (A4.4)$$

Where M is the mobility and c is the vacancy concentration. Since recovery influenced by the climbing process implies diffusion of vacancies or interstitials to the area of climb, this process is a thermal dependent process. This model is only considering point defects like vacancies and not interstitials since the activation energy for interstitials usually are 2-4 times higher than for vacancies, Ref. [A4.4]. The mobility is described as:

$$M = M_0 \cdot c \cdot D_m \cdot G \cdot b^3 \cdot \frac{1}{k \cdot T} \quad \dots (A4.5)$$

D_m , the vacancy migration times the vacancy concentration is representing the diffusion coefficient. G is the temperature dependent shear modulus, k is the Boltzmann's constant.

$$G = G_0 \cdot \left(1 - C \cdot \frac{T_k - 300}{1810}\right) \quad \dots (A4.6)$$

By combining Eqs. A4.1-A4.6 this gives us an equation for the derivative of the dislocation density.

$$\frac{d\rho}{dt} = \frac{d\rho_{gen}}{dt} - \frac{d\rho_{rec}}{dt} = \frac{m \cdot \dot{\epsilon}}{b} \cdot \left(\frac{1}{L_0 \cdot R_0} + \frac{\sqrt{\rho}}{k_L} \right) - M_0 \cdot c^2 \cdot \rho^2 \cdot D_m \cdot G \cdot b^3 \cdot \frac{1}{k \cdot T} \quad \dots (A4.7)$$

A formulation of the vacancy concentration in Eq. (A4.7) is needed since it plays an active role in the recovering part. The vacancy concentration could as well as the dislocation density equations be divided into two different parts; generation and recovery as follows:

$$\frac{dc}{dt} = \frac{dc_{gen}}{dt} - \frac{dc_{rec}}{dt} \quad \dots (A4.8)$$

Deformation of the material produces a considerable amount of vacancies due to among others interactions between dislocations.

$$\frac{dc_{gen}}{dt} = C \cdot b \cdot \dot{\epsilon} \cdot \sqrt{\rho} \quad \dots (A4.9)$$

C is a constant that needs to be optimised and the rest of the parameters are the same as described in the text above. Recovery is represented by diffusion of vacancies towards areas with high concentration of dislocations.

$$\frac{dc_{rec}}{dt} = C \cdot D_m \cdot c \cdot (c - c_0) \cdot \sqrt{\rho} \quad \dots (A4.10)$$

C is a constant and D_m times c is once again representing the diffusion coefficient for diffusion of vacancies and c_0 is the formulation of the equilibrium number of vacancies present in a material, see Eq. (A4.11). This equation is well known and widely used to predict the number of point defects in a material.

$$c_0 = \exp\left(-\frac{Q_v}{RT}\right) \quad \dots (A4.11)$$

Q_v is the activation energy for generation of vacancies; R is the universal gas constant and T the temperature. D_m , the vacancy migration used in Eqs. (A4.5), (A4.7) and (A4.10) is represented by Eq. (A4.11).

$$D_m = D_{m0} \exp\left(-\frac{Q_m}{RT}\right) \quad \dots (A4.12)$$

D_{m0} is the initial migration and Q_m is the activation energy for migration. Migration could be described as the rate at which a point defect moves from site to site in the lattice.

A4.2.2 Recrystallisation

Recovery is a collection of different processes whereas the climbing and cross slip phenomena's are two basic ones. The maybe most important one or at least a very powerful one is the recrystallisation which has been in the scope of researching for several decades. Since MEFOS by itself not have the deep material insight a cooperation group where put together in the spot of the project. The main objective was to develop the presented constitutive model to cover recrystallisation in multi pass situations and dynamic recrystallisation where the material has a mixed structure of deformed and recrystallised grains. It was started with a recrystallisation model developed at SIMR, Ref. [A4.2]. The model had been used for one-step deformation and static recrystallisation and was programmed in another computer language. It needed to be transformed into FORTRAN in order to fit the code format of the FEM program LS-DYNA. This work was performed but unfortunately it turned out that the model not was applicable at all for dynamic situations. When the situation were discovered the first model had to be reformulated and developed in order to cover situations with mixed grain sizes etc. The present model is based on the previous static recrystallisation model and was developed by SIMR, MIKRAB and MEFOS together. Worth to mention is that the equations in the recrystallisation model is prepared for effects influenced by precipitation.

A4.2.2.1 Subgrain

The whole concept was based on the subgrain structure development in deformed materials. Before we have achieved fully developed grains this stage is proceeded by formation of subgrains and the mean subgrain size can be approximated by Eq. (A4.13):

$$\bar{r}_{\text{subgrain}} = k_{\text{subgrain}} \cdot L \quad \dots (A4.13)$$

k_{subgrain} denotes a constant which has to be optimised and L is the mean free distance of dislocation migration. The growth rate for subgrains is described by:

$$\frac{d\bar{r}_{\text{subgrain}}}{dt} = \begin{cases} -k_{\text{subgrain}} \cdot \frac{L^2}{2k_L \sqrt{\rho_{\text{def}}}} \cdot \frac{d\rho_{\text{def}}}{dt} & (\dot{\epsilon} > 0) \\ k_{\text{sgg}} \cdot M_{\text{subgrain}} \cdot F_{\text{subgrain}} & (\dot{\epsilon} = 0) \end{cases} \quad \dots (A4.14)$$

where the two formulas is used weather there is an ongoing deformation or not. Basically during deformation the growth rate is dependent of the dislocation density and its generation rate in deformed areas. K_L is a constant and L is the mean free slip distance as denoted before but in this case it would maybe be represented as the mean free distance between obstacles in the matrix. When no deformation is present the subgrain growth is driven by reduction of subgrain boundary energy. The driving force is represented by Eq. (A4.15), where F_p is the pinning force of particles.

$$F_{\text{subgrain}} = \frac{\gamma_{\text{sg}}}{\bar{r}_{\text{sg}}} - k_{\text{ps}} F_p \quad \dots (A4.15)$$

A4.2.2.2 Nucleation

The classical nucleation theory can not be adopted for recrystallisation. Instead the description of nucleation in the present model is based on the abnormal subgrain growth mechanism. The critical subgrain size r_c , is represented by eq. A4.16 and can be described as the transition size from subgrain embryo into a growing new grain. r_c is far larger than the mean subgrain size. This would mean that only abnormal growing subgrains have the possibility to reach the critical size and the nucleation process becomes an abnormal process.

$$r_c = \frac{\gamma_g}{c_d G b^2 (\rho_{\text{def}} - \rho_0) - k_{pg} F_p} \quad \dots (A4.16)$$

γ_g is the grain boundary energy, ρ_{def} and ρ_0 is the dislocation density in deformed respectively the initial material state, c_d and k_{pg} are constants.

The abnormal growing subgrains must satisfy the size and disorientation conditions described further in Ref. [A4.3]. A critical subgrain size, r_{ac} is defined as the threshold for abnormal growth:

$$r_{ac} = k_{ac} \cdot \bar{r}_{\text{subgrain}} \quad \dots (A4.17)$$

Abnormal growth rate:

$$\frac{dr_a}{dt} = \begin{cases} 0 & (\dot{\epsilon} > 0) \\ k_{ra} \cdot M_{\text{subgrain}} \cdot F_{ra} & (\dot{\epsilon} = 0) \end{cases} \quad \dots (A4.18)$$

Where

$$F_{ra} = \frac{\bar{\gamma}_{\text{subgrain}}}{\bar{r}_{\text{subgrain}}} - \frac{\gamma_a}{r_a} \quad \dots (A4.19)$$

During deformation, subgrains generally not having time for abnormal growth, the new grains are only coming from the subgrains that are larger than the critical size, r_c , following relationship hold for the number of grains:

$$N = N_s \cdot \int_{r_c}^{\infty} p(r) dr \quad (\dot{\epsilon} > 0) \quad \dots (A4.20)$$

Where N_s is total site number and $p(r)$ is the size distribution function of the subgrains.

After the deformation, the nucleation is dependent on the number of abnormal growing subgrains and their growth rate.

$$\frac{dN_{\text{recg}}}{dt} = k_n \cdot N_s \cdot p(r_t) \cdot \left(\frac{dr_a}{dt} \right)_{r_a=r_c} \quad (\dot{\epsilon} = 0) \quad \dots (A4.21)$$

r_t is the initial size (at $t=0$) of the subgrains which reach r_c at time t :

$$r_t = r_c - \int_{t_0}^t \left(\frac{dr_a}{dt} \right) dt \quad \dots (A4.22)$$

A4.2.2.3 Growth and coarsening

The final grain size of recrystallised structure depends on both grain growth and coarsening. The growth and coarsening happen concurrently during recrystallisation. Here the growth means the grain growing in the deformed part, and the coarsening means grain coarsening inside the recrystallised part. In order to calculate the growth and coarsening simultaneously, recrystallised grains are treated as two classes – surface grains and interior grains. The surface grains grow to the deformed area. Reduction of storage energy is driving force and grain boundary plays as restrainer, described in Eqs. (A4.23) and (A4.24). The interior grains follow coarsening mechanism and the reduction of grain boundary is the driving force, see Eqs. (A4.25) and (A4.26). The grain growing rate combining growth and coarsening is given in Eq. (A4.27).

$$\frac{dR_{surf}}{dt} = k_{rg} M_g F_{recg} \quad \dots (A4.23)$$

$$F_{recg} = -\frac{\gamma_g}{R_{rec}} + c_d G b^2 (\rho_{def} - \rho_{rec}) - k_{ps} F_p \quad \dots (A4.24)$$

$$\frac{dR_{inte}}{dt} = k_{gg} M_g F_{grng} \quad \dots (A4.25)$$

$$F_{grng} = \frac{\gamma_g}{R_{rec}} - k_{pg} F_p \quad \dots (A4.26)$$

$$\frac{dR_{rec}}{dt} = \begin{cases} \frac{R_{rec}^2 \cdot \left(N_{inte} \cdot \frac{dR_{inte}}{dt} + N_{surf} \cdot \frac{dR_{surf}}{dt} \right) + \frac{r_c^3}{3} \cdot \frac{dN_{rec}}{dt}}{R_{rec}^2 \cdot \left(N_{rec} + \frac{dN_{rec}}{dt} \right)} & (f_{rec} < 1) \\ \frac{dR_{inte}}{dt} & (f_{rec} = 1) \end{cases} \quad \dots (A4.27)$$

Where N_{inte} and N_{surf} are grain numbers of interior and surface grains.

A4.2.2.4 Dynamic and Multi-step deformation

After the onset of recrystallisation, the material is separated into two parts: a recrystallised and non-recrystallised matrix. In case of dynamic recrystallisation, the deformation continues in both parts, but at different strain rates. The recrystallised (soft) part has a higher strain rate and the already deformed part has a lower strain rate. The grain sizes in these two parts are given as R_{rec} and R_{def} . The global strain rate is given as $\dot{\epsilon}$. The strain rate in these two parts can be calculated from Eqs. (A4.28) and (A4.29). In case of multi-step deformation, if the recrystallisation is not complete during the pause time, the situation is similar as dynamic recrystallisation case. The recrystallised and non-recrystallised materials will be deformed according to Eqs. (A4.28) and (A4.29). If the recrystallisation is

complete during pause time, then the deformation can be treated simply in the same way as if it was a one step deformation.

$$\dot{\epsilon}_{\text{rec}} = \left(E + \frac{D\dot{\epsilon}}{1-f_{\text{rec}}} \right) / \left(A_{\text{rec}} U_{\text{rec}} + \frac{Df_{\text{rec}}}{1-f_{\text{rec}}} \right) \quad \dots \text{(A4.28)}$$

$$\dot{\epsilon}_{\text{def}} = (\dot{\epsilon} - f_{\text{rec}} \dot{\epsilon}_{\text{rec}}) / (1 - f_{\text{rec}}) \quad \dots \text{(A4.29)}$$

where

$$D = m_{\text{rec}}^2 \alpha_{\text{rec}} G \frac{g_0}{2\sqrt{\rho_{\text{rec}}}} + m_{\text{def}}^2 \alpha_{\text{def}} G \frac{1}{2L_{\text{def}} \sqrt{\rho_{\text{def}}}} \quad \dots \text{(A4.30)}$$

$$E = \frac{1}{2} M c_{\text{rec}} m_{\text{rec}} \alpha_{\text{rec}} G b \rho_{\text{rec}}^{3/2} \left(1 - \frac{k_{\text{pr}} F_{\text{p}}}{\sqrt{\rho_{\text{rec}}}} \right) + \frac{k_{\text{gs}}}{2R_{\text{rec}}^{3/2}} \cdot \frac{dR_{\text{rec}}}{dt} \\ - \frac{1}{2} M c_{\text{def}} m_{\text{def}} \alpha_{\text{def}} G b \rho_{\text{def}}^{3/2} \left(1 - \frac{k_{\text{pr}} F_{\text{p}}}{\sqrt{\rho_{\text{def}}}} \right) - \frac{k_{\text{gs}}}{2R_{\text{def}}^{3/2}} \cdot \frac{dR_{\text{def}}}{dt} \quad \dots \text{(A4.31)}$$

$$A_{\text{rec}} = m_{\text{rec}} \alpha_{\text{rec}} G b \frac{1}{2\sqrt{\rho_{\text{rec}}}} \quad \dots \text{(A4.32)}$$

$$U_{\text{rec}} = \frac{m_{\text{rec}}}{(L_{\text{rec}} + g_0)b} \quad \dots \text{(A4.33)}$$

where

$$g_0 = \frac{4(1-f_{\text{rec}})}{R_{\text{rec}}} \quad \dots \text{(A4.34)}$$

A4.2.2.5 Flow stress

The influence of microstructure at the flow stress can be well described by:

$$\sigma = \sigma_0 + m \alpha G b \sqrt{\rho} \quad \dots \text{(A4.35)}$$

σ_0 is the initial frictional stress generally dependent of the temperature and the fraction of precipitations in the matrix, ρ is the mean dislocation density in one cell for partly recrystallised structures and the dislocation density for fully deformed or re-crystallised structures. α is a parameter used to describe the strength of obstacles, this parameter is treated as a constant independent of e.g. temperature and strain rate etc. The parameters m , G and b is the Taylor factor, shear modulus and Burgers vector respectively calculated and described as in the proceeding text. These parameters are slightly dependent of temperature and the strain level.

A4.3 Optimisation

Optimising in order to find extreme points of a function or systems of functions is a topic that by itself can be an objective of research. Optimisation generally becomes very complicated when the number of degree of freedoms are increasing and when the functions to be optimised are coupled and strongly dependent of each others. When optimising, maybe the hardest thing is to know weather or not the global minima is reached or if there is a local minima that has been found in the error function. The basic idea however here is to create a function F which for each set of the vector x returns a scalar which can be minimized regarding to a representative norm.

$$\min F(f_1(x_1, x_2 \dots x_n), f_2(x_1, x_2 \dots x_n), \dots, f_m(x_1, x_2 \dots x_n)), x \in \mathbb{R}^n \quad \dots (A4.36)$$

F is a function dependent of sub-functions, F is continuous and has continuous derivatives in the n dimensional space. This is essential since optimisation algorithms usually need calculations of the first and second order derivative. The functions f_1 to f_n are the set of first order ordinary differential equations representing the constitutive model presented in the proceeding section. The vector x are forming a multidimensional space where all elements are representing one axis and where the minima should be found. A lot of efforts have been put on the optimisation part since the parameters are important for the ability to predict the final flow stress. It is impossible to present all the attempts made within the project except for some essential steps used in the optimisation process.

A4.3.1 Screening

Screening means that the measured data which should be used for the optimisation has to be checked and/or cleaned from irrelevant samples which always are present. Almost in all cases there is a need for screening the data in order to extract as much information as it is possible from it. There are always scattering in measurement devices due to vibrations, human factor and samples are performed in another reason etc. where all factors, called secondary factors are influencing the result in an order that not are represented by physical reasons rather then by randomised factors.

A4.3.1.1 Oscillations

In this project the C-Mn-Nb steel data were gathered in the frame of this project but needed filtering algorithms in order to stabilize the strain rate which was oscillating heavily. When using the data in models which are directly dependent of the strain rate like the constitutive models in chapter 0 one is obtaining troubles if the amplitude in the strain rate is too high. Once again data is data and this amplitude might be realistic but for modelling purposes it can be more favourable e.g. to use the local mean value instead.

For this reason a method called the "Loess" method, Ref. [A4.5] were used in the smoothing process of the data. "Loess" are derived from the term "locally weighted scatter plot smooth" and is locally approximating data by linear regression, samples in the neighbourhood to the one to be approximated is weighted which is the width of the filter. However in Fig. A4.1, smoothed strain rate data is presented which clearly are pointing out the effectiveness of the filtering algorithm. Usually ten up to about fifty samples are used in the local regression model.

The problem with an oscillating strain rate is that when integrating the constitutive model with a specific time increment different from the sampled one. The constitutive models are

sensitive at the strain rate level whereas strong oscillation can influence the optimisation result.

A4.3.1.2 Interpolation

Data for the Cr steel was as mentioned in the Section 0 not produced for this project from the first time. Hence the number of samples was too few and it is not possible to see what history the data has, Fig. A4.2, if they have been smoothed, if the strain rate is corrected etc. However the easiest and most cost efficient way to expand the dataset is to interpolate new samples based on the original samples with the drawback that information in the extra samples are constructed and not real. The shape of the flow stress curve is though almost continuous and smooth so there is no major risk by using interpolation. In Fig. A4.2, the marked curve is representing the interpolated data which seems to behave in a reasonable way except for the end of the curve where the curvature seems to be slightly too high.

A4.3.2 Normalizing

When minimising F in Eq. (A4.55) by an algorithms that uses the gradient calculations it is logical to realise that if two directions or elements of the input vector x have magnitudes of 10^{15} and 10^{-5} like the dislocation density respectively the mean free slip distance, there is a risk that the error in the derivative calculations for one parameter is larger than the absolute value of another one. Therefore it could be necessary to normalise the input vector. An easy way to achieve this is to initially use an internal parameter vector usually equal to the unit vector outside the algorithm, inside the algorithm the elements of the parameter vector x , is transformed to accurate magnitudes by element wise multiplication with vector m . This means that:

$$\bar{x}_i = x_i \cdot m_i; i = 1 \dots n \quad \dots (A4.37)$$

Where n is the number of parameters to be optimised and x_i is the “scaled” parameter vector.

A4.3.3 Norm and optimisation criteria

As important as the rest of the optimisation is the level where one could say that the optimisation is successful and should be terminated. A convenient and very common error estimation to use is the variant of the L_2 norm called the RSM (Root Mean Squared) error or simply RSME:

$$RMSE = \left(\frac{1}{n} \sum_{i=1}^n (x - \hat{x})^2 \right)^{1/2} \quad \dots (A4.38)$$

x and $\hat{x}$ denotes the measured respective the predicted sample and n the number of samples. An advantage of using the RMSE is that it is so widely used that it easily can be related to similar work or that one is familiar with it.

Typical RMSE for optimisation of one strain rate at one temperature use to be in-between the range 0.5-1.5 MPa. For optimisation of one set of strain rates at one temperature the error has increased to about 4-10 MPa. When optimising a set of yield stress curves with various strain rates at one temperature the error for each of the strain rates is use to be approximately the same. When optimising both temperatures and strain rates the error is generally increasing since it becomes more difficult to find the true global minima. Usually

the optimisation error is dominated and very sensible by the predictions at lower strains basically because of two reasons. First is of a direct kind which means that the prediction error is large when the derivative is high and the second one is that if the initial derivative are wrongly predicted this influences the rest of the integration of the constitutive models since there are differential equations.

A4.3.4 Algorithms, strategies and results

There are a numerous number of different optimising algorithms widely used and available in both literature and as computer code. Which one who will produce the best optimising result is generally a very hard question to answer. And in many cases it will need a great knowledge about functional behaviour and algorithms in order to choose the optimal one. Generally a trail and error procedure is the only way to find out the best method for the specific problem. A broad categorisation of methods would be, methods that only use function evaluations like the Nelder-Mead's Simplex method, gradient based methods like "Steepest descent" and higher order derivative methods such as the Newton's method. Newton based algorithms usually need the Hessian matrix to be calculated which is very time- and memory consuming process. Quasi-Newton methods are a modification of these and are avoiding this drawback by updating the Hessian matrix and are generally much more efficient, example of algorithms are the well known Levenberg-Marquardts method and BFGS, Ref. [A4.6]. What can be said though is that the RAM memory usage differs a lot and should be a criterion itself when choosing an algorithm. For optimisation of the parameters in these constitutive models all kind of optimisation algorithms has been investigated but the Levenberg-Marquardts method with a line search method has generally produced the best results.

What was discussed above is one aspect of the complete optimising problem, which however not is fully highlighted yet since the way in applying the algorithms has its own important role and can be varied in numerous ways.

The stress level in the flow stress data is dependent of three major variables, the strain rate, the temperature and the strain level. In Section 2 there were discussed about recovery mechanisms such as climbing and cross slip. Climbing is one of the processes that mainly are dependent of the temperature instead of the strain rate since the way a dislocation is climbing is due to thermal activated diffusion of a vacancies or interstitials. This is pointing out that the optimisation itself could be separated to cover the temperature or/and the strain rate dependentness each by each or all together. In the project a lot of different strategies have been applied where among the most successful ones has been to first optimise for a medium temperature let say 900°C. This is tuning the constants into a reasonable level whereas the optimisation can be extended with strain rates at another temperature or temperatures, assumed of course that the first optimisation were successful.

Since a material response are dependent of many primary and secondary effects which has their strongest significance at different temperatures, strains or strain rates it is generally difficult to optimise the parameters for all situations. When the optimised parameters have been used in FEM simulations usually the parameters have been optimised for two or three temperatures and for all strain rates. The important thing then is that the temperature span in the optimisation is covering the span that will be present in the simulations. One should though remember that two primary effects that have a great significance are recrystallisation and precipitation which not have been included in the optimisations.

A4.3.5 Optimisation of Cr steels

A4.3.5.1 Polynomial prediction of the initial frictional stress

For all flow stress data, the most difficult part to optimise except for the recrystallisation part of the curve is the one at strains lower than 0.05. One simple reason is that there was no information available for this particular Cr data, see Section 0. Other indirect reasons are that there are high derivatives in this area and that the dislocation density is increasing so fast that even small deviations in optimised parameters have great effects at the shape of the rest of the curve.

To overcome the missing of samples in the low strain regions, optimisation were just applied to the data that existed at strains greater than 0.05. A polynomial function, Eq. (A4.58), was fitted to the sampled stress levels at strains equal to 0.05 for all temperatures and strain rates:

$$\sigma_{0.05} = f(\dot{\epsilon}, T), \epsilon = 0.05 \quad \dots (A4.39)$$

For stress calculations at strains lower than 0.05 a linearly interpolated stress level based on the predicted one could be used, which is shown in Fig. A4.3. The polynomial predictions are very accurate relatively and coefficients in the polynomial function were easily fitted to the measured stresses at strains equal to 0.05. In Fig. A4.4 this method is shown whereas predictions at different temperatures and strain rates are plotted in the same figure. The black curve is showing a prediction at 750 C which is an extrapolation of the constitutive model, the red curves are interpolations and the blue ones are predictions at data that the model already has been exposed to. The figure is showing that the curves not have continuous derivatives which can have effect on the convergence of FEM solution. The prediction at 750°C are actually not relevant since the temperature is too low but it is very important to expose the model for extrapolations in order to get a feeling for the robustness of it. In Fig. A4.5, the extrapolation of the strain rate is plotted. Also in this figure there is no strange behaviour of the curve outside of the strain rate span used in the optimisation.

A4.3.5.2 Temperature dependent initial frictional stress

The method applied in the proceeding section is possible to implement, it gives reasonable results and the major advantage is that it is easy to find a global minima that cover all temperatures and strain rates with the same parameters. The drawback though is that it is not accurate in a physical meaning to assume that all curves should have the same dislocation density at strains equal to 0.05 but different stresses. In fact the only or almost the only variable of greater significance in the flow stress equation, Section 0 is the dislocation density. A more physically accurate representation then is to assume that the initial frictional stress at strains equal to zero is a thermally dependent function like:

$$\sigma_0 = f(T), \epsilon = 0 \quad \dots (A4.40)$$

Since data for strains lower than 0.05 was not available for the Cr steel, two alternatives were possible. First, one could approximate this part quite reasonably by just construct samples that are maintaining the shape of the curves. Secondly one could produce a sampling sequence where these parts are measured. The first alternative was adopted by financial and logical reasons. However, by extending the curves to zero strains and approximating a reasonable initial frictional stress the optimisation becomes more difficult. In Fig. A4.6 predictions for strain rates at 800° is made where the RMSE is approximately 1.63 MPa. When optimising for several temperatures the RMSE generally increases, once

again because the models has to cover a wider range of effects. In Fig. A4.7 and Fig. A4.8 optimising results for two and three different temperatures are presented. RMSE's in the figure are 4.2 MPa respectively 6.78 MPa.

A4.3.6 Optimisation of C-Mn-Nb steels

The flow stress measurements of the C-Mn-Nb steels were performed by TU Freiberg. For optimisation, flow stresses at three strain rates, 0.1, 1 and 10/s was logged and has been used in the optimisation together with five different temperatures 800, 850, 900, 1000, 1100 and 1200 °C. The strain rates does not cover the full range of strain rates present in the benchmark rolling case but it cover the most important region where the flow stress behaviour changing at most. If the optimised parameters are physically reasonable the models have showed a stable behaviour when extrapolating.

Notable for optimisation of these data is that no use of the polynomial prediction of the initial frictional stress has been used. This is due to that data for lower strains are present in the measurements and that it is not a physically accurate interpretation.

The optimisation for the C-Mn-Nb data has been much more difficult than optimisation of Cr steel data. There is unclear what the reason for this is but one reason might be that the highest strain rates sometimes at several temperatures are showing a peculiar shape at low strains which could come from troubles with the measurement device. Another reason could be the presence of particles but this would be case for the Cr steel for the Cr steels as well. However, different optimisation techniques and algorithms have been used where the result has been approximately the same. Figures A4.9 to A4.13 are showing results for optimisation at temperatures from 850°C up to 1200°C where the typical RMSE lies around 4-8 MPa. When looking at Figs. A4.9 to A4.13 it is obvious that the dynamic recrystallisation not is represented by any equations in the constitutive model. Since the recrystallisation model not has been fully implemented into the rest of the constitutive models the optimisation should just concern fitting of parameters in the vacancy concentration- and dislocation generation equations. Figure A4.14 is showing the result for optimisation of flow stresses at two temperatures and at three strain rates.

A4.4 Implementation

In the former project, Ref. [A4.1] a subroutine were also implemented into LS-DYNA for applying the constitutive model to FEM simulations. A lot of things has though been changed since then e.g. that the FEM program has changed, the constitutive model has been developed in some areas, and optimisation is improved.

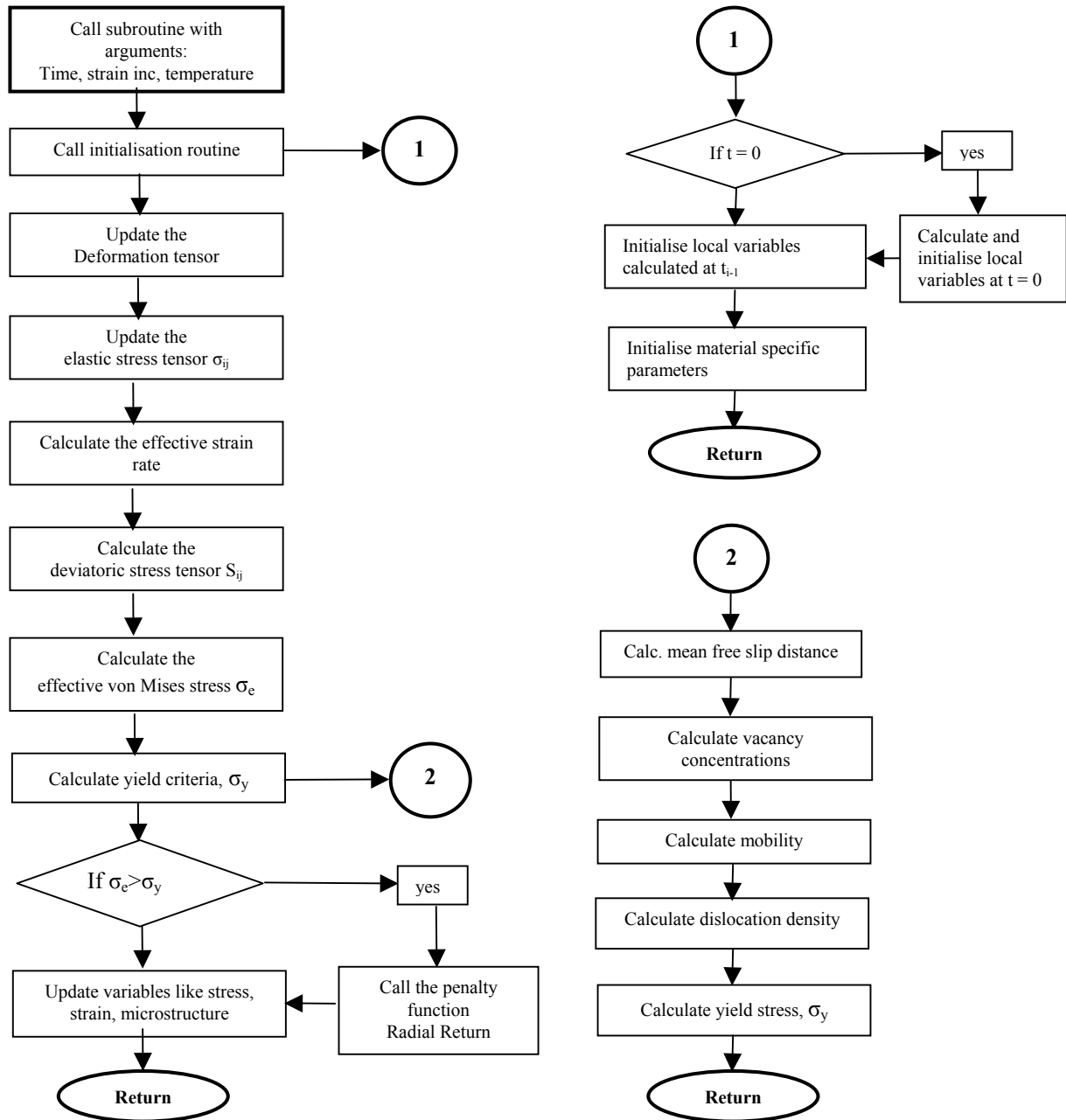
The implementation has been compiled only with object files for the single processor version which means that the solution times could be decreased if they were a problem, for small simulations like the one made in this project this however does not matters.

A4.4.1 Explanation of subroutine

In Section 0, where the constitutive model is presented it can be seen that the governing microstructure equations are a set of first order ordinary differential equations. To solve these they must be integrated in respect of time. This can be achieved by some advanced algorithm like e.g. the well known Runge-Kutta method or if the time increment is very short by a direct approximation like the Euler method where the derivative is multiplied by the time increment as in Eq. (A4.41).

$$y_{n+1} = y_n + \Delta t \cdot \frac{dy}{dt} \quad \dots (A4.41)$$

Since LS-DYNA is based on an explicit FEM method which uses very short time increments it is very convenient to use fast Euler integration schemes for the subroutine without reduced accuracy in the calculations.



Flowchart A4.1: schematic figure for implementation of the UMAT (User Defined Material Subroutine)

In Flowchart A4.1, a schematic description of the implemented UMAT subroutine is described. It works like the subroutine is called at every time increment with the time, strain,

old stresses, time increment history variables and material specific constants as arguments. In the beginning of the program a subroutine to the main program is initialising history variables and material specific parameters and at the first time increment the initial values also has to be calculated, marked as sequence 1 in Flowchart A4.1.

Based on calculated incremental deformations and strains in the main program first step in the UMAT is to update the stress tensor for an assumed fully elastic deformation.

A4.4.1.1 Basic equations

Dependent of what kind of FEM formulations that are used the incompressibility is treated in different ways. In LS-DYNA the incremental stress is represented by Eq. (A4.42).

$$\sigma_{ij}^n = \kappa \cdot \varepsilon_v^n \cdot \delta_{ij} + 2 \cdot G \cdot \bar{\varepsilon}_{ij}^n, i = 1..3, j = 1..3 \quad \dots (A4.42)$$

Where δ is the Kronecker delta and κ is the bulk modulus

$$\kappa = \frac{E}{3(1 - 2\nu)} \quad \dots (A4.43)$$

G is the shear modulus:

$$G = \frac{E}{2(1 + \nu)} \quad \dots (A4.44)$$

and $\bar{\varepsilon}_{ij}^n$ in Eq. (A4.42) is the deviatoric strain components used in order to neglect effects influenced by the hydrostatic stress.

$$\bar{\varepsilon}_{ij}^n = \varepsilon_{ij}^n - \frac{1}{3} \varepsilon_v^n \delta_{ij}, i = 1..3, j = 1..3 \quad \dots (A4.45)$$

And finally the ε_v^n is the volumetric strain which is anticipated to be small in comparison with the deviatoric strains since the material in scope are almost incompressible.

$$\varepsilon_v^n = \varepsilon_{kk}^n, k = 1, 2, 3 \quad \dots (A4.46)$$

By Eqs. (A4.42) to (A4.46) and the stress at t_{n-1} the updated stress is represented by following relationship:

$$\sigma_{ij}^n = \sigma_{ij}^{n-1} + \kappa \cdot \varepsilon_v^n \cdot \delta_{ij} + 2 \cdot G \cdot \bar{\varepsilon}_{ij}^n, i = 1..3, j = 1..3 \quad \dots (A4.47)$$

The updated stress in Eq. (A4.47) is a first elastic trial where it is assumed that all deformations are producing only elastic responses.

Updating the incremental effective strain rate is performed by Eq. (A4.48):

$$\dot{\varepsilon}_e^n = \sqrt{\dot{\varepsilon}_{xx}^2 + \dot{\varepsilon}_{yy}^2 + \dot{\varepsilon}_{zz}^2 + 2 \cdot \dot{\varepsilon}_{xy}^2 + 2 \cdot \dot{\varepsilon}_{xz}^2 + 2 \cdot \dot{\varepsilon}_{yz}^2} \quad \dots (A4.48)$$

A4.4.1.1.1 Plasticity

In order to be able to describe the plastic behaviour of an elastoplastic material when it is deformed, it is essential to have a representative yield function of the form.

$$f_y(\sigma_{ij}, \varepsilon_{ij}^p, \dots) \quad \dots \text{ (A4.49)}$$

Where the dots do denotes material specific state variables like the dislocation density or any other variable describing the microstructure which influences the yield surface. The material response will be pure elastic as long as the yield function is lower then zero.

$$f_y < 0 \quad \dots \text{ (A4.50)}$$

and elastic or plastic when the function is equal to zero.

$$f_y = 0 \quad \dots \text{ (A4.51)}$$

Notable is that positive values of the yield function are inadmissible. The function f is usually implemented as

$$f_y = \sigma_e^2 - \sigma_y^2 \quad \dots \text{ (A4.52)}$$

Where σ_e is the effective elastic stress according to von Mises plasticity theory and σ_y is the yield stress calculated by the constitutive model.

$$\sigma_e = \sqrt{\frac{3}{2} \mathbf{S}_{ij} \cdot \mathbf{S}_{ij}} \quad \dots \text{ (A4.53)}$$

$\mathbf{S}_{ij}$ is the deviatoric stress tensor calculated by

$$\mathbf{S}_{ij} = \sigma_{ij} - \frac{1}{3} \sigma_{kk} \cdot \delta_{ij}, i, j, k = 1..3 \quad \dots \text{ (A4.54)}$$

In order to check if yielding occurs or not the only step that last in the subroutine is to calculate the yield stress which is done by the constitutive models and Eq. (A4.35), the modified Hirsch equation. This is marked as sequence 2 in Flowchart A4.1. If the statement in Eq. (A4.50) is true the program will adopt the elastic trial calculations and save the incremental stresses as history variables. If the statement in Eq. (A4.50) not is fulfilled the material response will be plastic and/or elastic. The yield function, Eq. (A4.51) still is the governing equation and the overshooting stress has to be transformed back to the yield surface by a penalty algorithm. There are several ones to use with a different level of complexity, in this subroutine however the well known Radial return method, Ref. [A4.7], were implemented because of its simplicity and reasonable accuracy.

A4.4.1.1.2 Radial return

The purpose with the algorithm is to radially scale the overshooting deviatoric stresses back to the flow surface defined by Eqs. (A4.51) and (A4.52). Routines like the Radial return method is not automatically implemented in LS-DYNA's source code, at least there is no easy way to call existing routines and this has to be manually implemented by the user. First step is to convert the overshooting stress to a plastic strain increment which is done by Eq. (A4.54).

$$\Delta \varepsilon_p = \frac{2}{3} \cdot \frac{(1+\nu) \cdot (\sigma_e - \sigma_y)}{E} \quad \dots (A4.55)$$

The plastic strain contribution has to be added and stored into a specific history variable devoted for the effective plastic strain. Next step is to calculate a penalty factor lambda to scale the deviatoric stress tensor back to flow surface.

$$\lambda = \frac{3}{2} \cdot \frac{E}{(1+\nu)} \cdot \frac{\Delta \varepsilon_p}{\sigma_e} \quad \dots (A4.56)$$

The last step is to subtract the converted plastic strain from the elastic stress case

$$\sigma_{ij} = \sigma_{ij} - \lambda \cdot S_{ij}, i, j = 1..3 \quad \dots (A4.57)$$

A4.4.2 Troubleshooting of compiling

In LS-DYNA the common way of extending the code with your own subroutines, like a new material subroutine is to write your code in FORTRAN and compile/link it together with the compiled source code in form of object files delivered by the software supplier. It could be a lot of troubleshooting with this since there is no or at least a very poor explanation of what parameters that should be used or what parameter that are public etc. Another thing that can make this hard to work with is that there are no or limited possibilities to debug the code.

In later part of 2004 the computer used for simulations at MEFOS were replaced by a PC cluster, this implied that LS-DYNA had to be updated and reinstalled in order to run on the new platform which was different from the former one. Since there not exists any options to compile and link subroutine and object files in a running mode, this caused big troubles to find a working compiler. The problem was to find a compiler that could run on the desired platform compile and link the subroutine with the object files libraries delivered by the supplier. Usually the suppliers use to have a number of versions of object files compiled with different compilers in order to be suitable for the customer's system requirements. In this case MEFOS could not find a compiler that could run on MEFOS platform and in the same time be compatible with the delivered object libraries. Finally, by contact with the Livermore Software Technology Corporation in USA one solution were find and could be installed at our system. This did though delay the planned work and unfortunately brought several weeks of extra work at problems not significant for the project.

A4.5 FEM Simulation

A4.5.1 Introduction

Two different simulation models have been set up during the project, the first one is a model for simulating compressions of circular test specimens and the second one is a model for simulating hot rolling. In the beginning of the project a benchmark example was worked out by the partners in order to have one reference point when discussing models etc. The benchmark model is based on geometries in the pilot plant at TU Freibergs workshop. Since MEFOS did not have any mill data available in the beginning of the project, the rolling model were just implemented but not used whereas the compression model was created for validation of the constitutive model against the measured data.

Major options that the FEM models were going to handle were:

- Thermo mechanical simulations: this includes radiation, convection and conduction
- Pause time simulation
- Reasonable short simulation times
- Contact, both thermal and mechanical

Both models, compression and benchmark, has been used for both thermo mechanical and only structural simulations. There is though an increase in simulation time when solving the relevant heat equations. Pause time simulations were not fully developed since it showed to be a quite time consuming work. The reason for this was that LS-DYNA is based on explicit FEM and explicit solutions is very seldom longer then a couple of seconds since the time steps are very short which means that the simulation times becomes very long if the pause times considered is about 5-10 minutes. This is though dependent of the number of elements, single or multi processor mode, or in what extent the simulation could be mass scaled.

Generally the thermo mechanical simulation possibilities have been improved in LS-DYNA but are still not flexible enough. E.g. when simulating pause times it would be very convenient to manually extend the time step and “turn off” the structural solver and only use the thermal solver which have longer time increments. This is impossible today when a user defined subroutine is compiled to the code.

When the rolling model were created, thermal conduction in use with the contact algorithm was very inefficient in searching for elements in contact. Due to this, the conduction option was put to a side and the most of the thermal modelling was performed with radiative and convective boundary conditions. This has effects on the solutions since it usually is a temperature difference between the roll and work piece of 8-900°C but it was chosen that the increased simulation time not was worth the small improvement in the thermal solution that is present since the simulation times are so short.

A4.5.2 Compression model

A4.5.2.1 Geometry, Boundary conditions and specification of simulation

The compression model was set up since it was convenient to be able to make fast simulations for validation of relatively simple cases and geometries. The compression models also are reflecting the way the data has been sampled whereas this is a good tool to make sure that models are implemented correctly and that the optimisation is accurate. For a long time there were only data available for compression tests and therefore it was needed to create a tool for validation of these particular yield stress data.

The model has been used for both mechanically and thermo-mechanically coupled simulations and is simulating the constant or varying strain rate compressions of circular specimens up to strains equal to 1, see Fig. A4.15.

In order to reduce the number of nodes and elements symmetry is used whereas only one eighth of the original specimen is modelled. Nodes on the bottom surface have been constrained to move in the axial or y-direction and are free to move in the x and z directions. Nodes on the inward surfaces are constrained in the x and y directions respectively in order to maintain the position of the cut. Boundary conditions for the top surfaces have been modelled in several different ways depending if contact should be included or not.

With contact included the stress state will act in a slightly different way since the friction is “holding” the upper surface. There is no easy work to predict an accurate magnitude of the friction coefficient which not has been of priority within this work where a coulomb friction model has been used where the friction coefficient typically has been put to 0.3. The forging dies have been modelled as rigid walls which mean that the dies not are deformable which also are affecting the results.

When contact not has been used, the set of nodes forming the upper surface has instead been given a prescribed motion and are free to move in the x-y plane. This is more easily to implement and there is less parameters that has to be prescribed for e.g. friction and penetration parameters. Without contact no un-symmetric thickening effect takes place which has influence at the effective stress state during the compression. This is however considered as minor effects and could be considered in order to improve predictions.

A4.5.2.2 Results and discussion

Simulations with the compression model are very fast even though there is a physically based material subroutine that has to be integrated at every time increment. For compression simulations up to a strain equal to 1 and at 1/s in strain rate, the simulation time is about ten minutes and for lower or higher strain rates the simulation time increases respectively decreases. The simulation time is dependent of the mass scaling, which in all simulation has been about ten to thirty times the original density. Validation of the model has been done at several temperatures and in Figs. A4.17 to A4.21 predictions of the effective yield stresses for the Cr steel at 900°C and at different strain rates are presented. The figures are presenting three different curves, the blue curve is showing the measured data, the red dashed one is the resulting yield stress from the FEM simulation and finally the black curve is representing the resultant curve when optimising the constitutive model. Figures A4.21 and A4.22 are showing the same kind of results but for the C-Mn-Nb. It can be seen that the results are accurate but with some deviances where the yield stress curves are flattening out or where the change in derivative is largest. In Figs. A4.16, A4.17 and A4.18 recrystallisation are present as can be seen at the dramatic softening of the curves

around strains equal to 0.4. In this figures it becomes clear that no constitutive model are dealing with this phenomena since the predictions are still maintaining an almost constant stress level. The predictions of the yield stresses of the Cr-Mo-V steel generally looks better but that might be effects from the construction of the initial frictional stress that might be too high whereas the changes in derivative are easier to predict.

In Figs. A4.22 to A4.25 results for the history variables for C-Mn-Nb steel are plotted. These figures are showing a simulation where contact is included and the resultant effective stress and effective plastic strain has both their maximum at the centre of the specimen. In this figures the effective plastic strain level is approximately 0.5 which is half of the strain maximum in the simulation. The dislocation density, plotted in Fig. A4.22 has at this stress level increased from an initial level of $10^{13}/\text{m}^2$ up to a level in the magnitude of $10^{15}/\text{m}^2$. If this are reasonable or not is hard to answer since it depends on different factors and must be validated by itself in order to ensure that this is the case. This is of course a risk when validation of the prediction ability of the constitutive model is based on just one parameter like the yield stress as in this case. The prediction of the yield stress can be very good whereas the microstructure parameters like the dislocation density not are physical at all. One way to get around this problem is to put constraints at some parameters when optimising the model and relate the constraints on measurements like for instance measurements of the dislocation density. However, Fig. A4.22 is showing that the dislocation goes hand in hand with the effective stress level as it should do with Eq. (A4.34) in mind. In Fig. A4.25 the vacancy concentration is plotted.

In Section 0 it was mentioned that one attempt to get around the missing of initial data was to predict the initial frictional stress and it was mentioned that a problem was that small deviations in the strain rate is affecting the prediction of the $\sigma_{0.05}$ stress level. In Fig. A4.26 the drawback with this formulation is obvious for a FEM simulation for the Cr-Mo-V steel.

A4.5.3 Rolling model

A4.5.3.1 Geometry, Boundary conditions and specification of simulation

In order to be able to compare the different partners a unified model were set up as a benchmark example as mentioned earlier. This benchmark example is a hot rolling case of a work piece with geometrical and technical prescriptions. The same case was also set up in TU Freiberg pilot plant and milling was performed in pilot scale at that environment.

This model has as the compression model also the thermo mechanical coupling condition available where simulations have been performed with both options. The constitutive model always needs a temperature but this can be prescribed without any solution of the temperature. When thermo mechanically coupled simulations has been performed the heat fluxes from the material to the surrounding takes place by radiation and convection. The conductive heat loss from the material to the work roll is the major surface heat flux together with the radiation in rolling situations. There is unfortunately no possibility to use the conductive contact option when one is using contact entities as models of the work rolls which means that one has to model it as rigid body. The problem formulation that occurs is that if the extended simulation time due to the increased number of elements is worth the slightly more accurate temperature calculation. For surface conditions the answer would be yes but for the overall result for the material behaviour it is discussible and was discarded in this work.

As was clarified in the discussion above was that the work roll has been modelled as a contact entity which means that it is described as a mathematical surface used for defining

contact, they are not considered as deformable bodies whereas no flattening of the rolls are considered in the model.

For simplicity reasons the model is split up into four quarters where just one are modelled because of the symmetry. Figure A4.27 is showing the geometric model where the small triad is indicating the positive directions of the model. All surfaces with a negative normal at the work piece are modelled as symmetry surfaces. The upper work roll which is the one that has been included due to the symmetry conditions are showed where it is obvious that it actually does not consists of any elements. The major advantage by using the entity option is that it decreases the simulation times due to the lower number of elements.

The friction force between the roll and the work piece is modelled by a coulomb friction model and no deformation of the roll takes place.

A4.5.3.2 Results and discussion

A number of different simulations have been performed within the project. But for the benchmark rolling simulations only the C-Mn-Nb steel has been of interest since this is the actual connection to the other partners and the natural point for discussing and comparing the models.

In Figs. A4.27 to A4.31 results from one of the simulations are shown. The simulation times in the figures are 0.3 seconds which means that the pass not are completed yet and in the rear part of the work piece there are still an ongoing deformation.

Figure A4.29 is showing the calculated yield stress or the effective stress which is seen to have its maximum somewhere in the deformation zone produced by the work roll. Relaxation becomes present in the positive x direction which is indicated as blue coloured areas and effects of the higher heat fluxes at the edges and corners could clearly be seen. In Fig. A4.32 the flow stress is plotted for two nodes where one is located in the x wise midsection and the other one is located in the front section of the work piece. The red curve is showing the stress level in the mid section and the stress is increasing from the start time even when this element not has been exposed to any deformation yet. The reason why the dislocation density and flow stress is increasing is that the material is experiencing an increased strain rate level because of the deformation introduced shock waves that are travelling forth and back through the material, see Fig. A4.34. The waves which are a result of the first contact between the work piece and the roll are only producing elastic distortions but they still have effects on the stress state whereas the flow stress are increasing onto it reaches the roll gap when time is approximately 0.17 s. When the work piece has entered the roll gap, flow stress is dramatically increasing up to its maximum level at approximately 190 MPa and the following relaxation is reducing the stress by 15% to approximately 160 MPa. In Fig. A4.34 the effective strain rate for one element in the mid section of the plate is plotted, it is showing that this mill operation are producing strain rates up to approximately 18/s.

In Fig. A4.28 the effective plastic strain are plotted, and there are some deviations in the work piece. Since the effective plastic strain is considered, no direct direction dependentness could be detected. The corner and the edge zones are experiencing a slightly lower effective strain because it not is increasing as much in length as the centre zone. In Fig. A4.33 the effective plastic strains are presented at the two elements as the flow stress presented for. Notable are that the strain level are slightly higher in the mid then in the front section.

Figure A4.31 are showing the calculated temperature profile and since the rolling time is only 0.35 seconds there are not enough time for any larger temperatures changes. The small temperature changing that is occurring is that all outward surfaces are slightly cooled and the corners and edges are as expected the zones with the highest heat loss showing a decreased temperature which influences the microstructure and resulting in a higher flow stress then the warmer kernel of the plate. In this figure heat conduction between the work roll and the test piece to the roll has not been included in the simulation.

Figure A4.30 is showing the vacancy concentration which remains almost constant when material has passed the deformation zone.

A4.6 Discussions

Physical models for describing the microstructure connected to material properties is an expanding part of the modelling area. Reason for this is that results can be used for a more accurate prediction of materials mechanical properties like the yield stress, fracture stress and elongation which can be used in process control situations. The underlying physical theories describing material behaviour like dislocation, recrystallisation and recovery is outlined for a several decades ago but of practical reasons not applied to reality before the computer science made this possible during the last decade. Today it is possible to run extensive and complicated models describing at lots of phenomena's without any increase in computational times compared to a situation for 10-15 years ago.

The perfect constitutive model is a model that can predict the material response under environmental influences like the temperature, deformation, chemical components etc and do perfect predictions of the microstructure which in turn gives the relevant mechanical property like the yield stress. This will of course never be the situation but one can continuously refine the prediction ability by adding reasonable physically based equation to the constitutive model. The model is today covering microstructure properties like the dislocation density and the vacancy concentration and their active role in the recovery and generation processes. Recoveries in the model are based on classical theories for thermal/diffusion controlled climbing and glide processes. Two major parts are still missing in the model which is the description of precipitations and recrystallisation. Since precipitation models generally need information about the chemical equilibrium in the material it will imply that these are calculated or predicted. During the project it were decided that it not was reasonable to put efforts in this work within this project. The recrystallisation model is developed in cooperation between MEFOS, SIMR and MIKRAB and are working for dynamic recrystallisation situations and when the matrix consists of mixed grain size structures; deformed and undeformed. The recrystallisation model are integrated with the original constitutive model and can be used for predictions but has not been implemented into LS-DYNA because a considerable time had to spent on compilation trouble and that the recrystallisation model had to be reformulated.

The constitutive model is working and seems to handle the most relevant microstructure properties and their evolutions but this necessarily does not mean that it predicts all microstructure variables as good as it predicts the flow stress. To be sure that this is the case one has to verify e.g. the calculated dislocation densities with measured ones, this is not made within this project but by experience from literature and earlier projects reasonable validation still can be made.

One very important part of the project is the optimising of the parameters in the constitutive model which also has been very time consuming. If we consider that the equations in the models are fully describing the physics of the microstructure and that there are no

measurement errors in the data present it would be easier to find a global minimum of the optimisation. That is not the situation and there is very hard to know if the global minimum is reached or if optimisation could be further improved since there will always be secondary effects that have influence in the final solution. Another aspect of the optimisation is that there are hard to know if the constitutive model is holding enough degrees of freedoms. If the constitutive model has too many degrees of freedoms the risk is that the model will be extremely dangerous to extrapolate outside the “input space” with respect to strain rate and temperatures. On the other hand, if the degrees of freedoms are too less the predictions will not be able to follow curvatures as well, which can be thought of a straight line which has a low degree of freedom. Dependent on in what situation the constitutive model will be used the input space in temperature for the optimisation should be in proportion to the problem temperature. This is a balance between the prediction ability which increases for a narrow temperature span and the temperature span needed in the simulations.

The implementation of the UMAT was made with help from the old project but there were however some changes made since the old UMAT were implemented in another version and on another computer platform. There is though not a straight forward work to implement a UMAT into LS-DYNA and without support there is impossible to know the name of different variables needed for calculations of the yield stresses. One could also expect that in order to handle plasticity, radial return would be an already programmed subroutine that could be called which is probably is but unfortunately not is public. However, the implementation of recrystallisation model was delayed partly because the change of computer system at MEFOS where the compiler and operative system had to be changed and caused problems in finding a combination that were working our tasks. Another reason for delay was that the recrystallisation model had to be developed and reprogrammed since the original one could not handle the mixed grain size structures in a reasonable way.

When the complete constitutive model was implemented, two kinds of deformation test cases were implemented; one for simulating a compression case and the other one for a benchmark rolling case. Simulations have been performed in different complexities with contact, thermal coupling options enabled in some cases. The main objective though is to check that the constitutive model is working reasonable and not to put a lot of efforts in improving for e.g. the friction or some other simulation specific parameter. The objectives of the simulations have been that they should be less detailed but detailed enough to catch the overall problem formulation. Results from simulation are showing a good agreement with measured data and when formulations of the initial frictional stress were reformulated the convergence in the solution procedure became even more stable.

A4.7 References

- A4.1 Ingham, P M: 'The effect of strain reversal and strain-time path on constitutive relationships for metal rolling/forming processes', Final Report, ECSC Project, 7210.EC/811, 2000, ISBN-92-894-1546-0, pp99-139.
- A4.2 A physical model for prediction of microstructure evolution during thermomechanical processing, Wang, X; Siwecki, T; Engberg, G, Materials Science Forum. 2003, Vol. 426-432, pp3801-3806.
- A4.3 Humphreys F.J., Acta mater., 45, (1997), 4231-4240.
- A4.4 *Theory of dislocations*, Hirth, J. P; Lothe J, second edition, ISBN 0-471-09125-1.

- A4.5 *Robust locally-weighted regression and smoothing scatterplots*, Cleveland, W. S. (1979), Journal of the American Statistical Association, 74, 829-836.
- A4.6 Optimisation toolbox, User's guide, version 2, Coleman, T; Branch, M.A; Grace, A.
- A4.7 *Calculation of Elastic-Plastic Flow, Methods in computational physics*, Wilkins, M. L; Alder, B; Fernbach, S; Rotenberg, M. pp211-263, Academic press, New York, 1964.
- A4.8 *LS-DYNA, Keyword user's manual*, version 960, Volume 1, 2001.

Figures

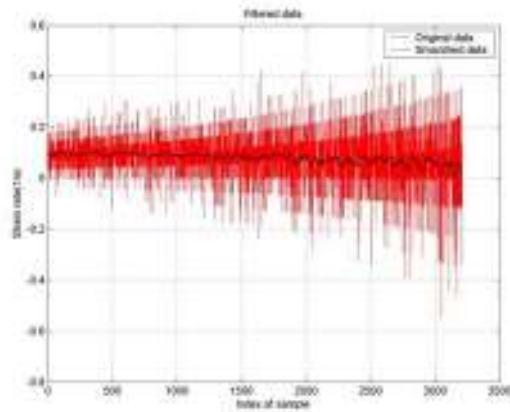


Fig. A4.1: Oscillations in strain rate

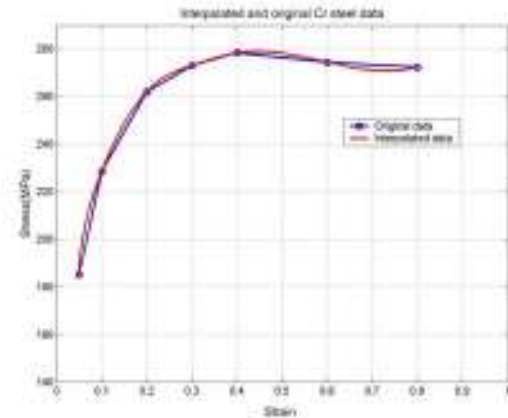


Fig. A4.2: Typical Cr steel data sampled at SIMR

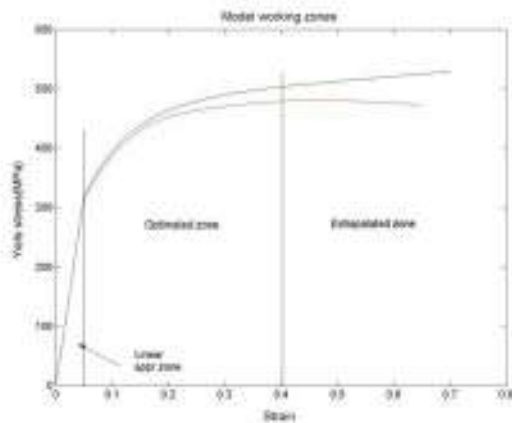


Fig. A4.3: Descriptive figure to show the use of initial frictional stress polynomial predictions

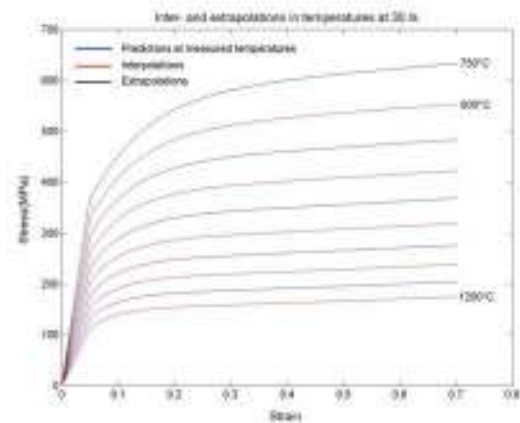


Fig. A4.4: Optimised flow stress curves with interpolated initial parts

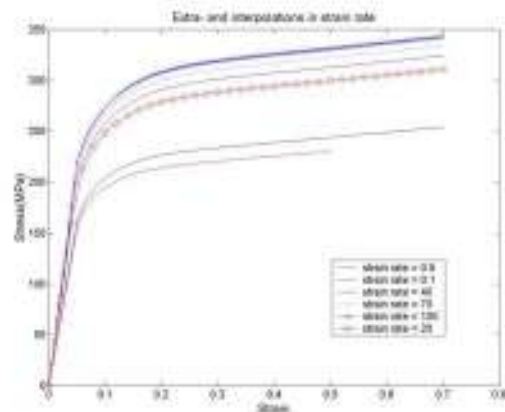


Fig. A4.5: Extrapolation of the strain rate

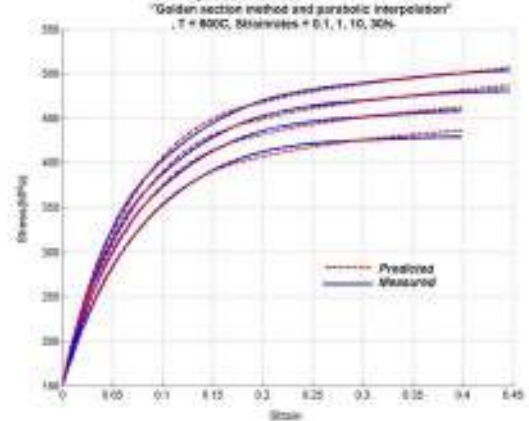


Fig. A4.6: Optimised flow stress curves with temperature dependent initial frictional stress

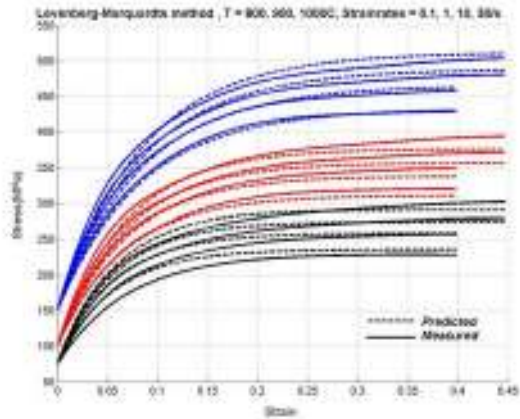


Fig. A4.7: Optimisation for Cr steel data at three different temperatures and four strain rates at each temperature

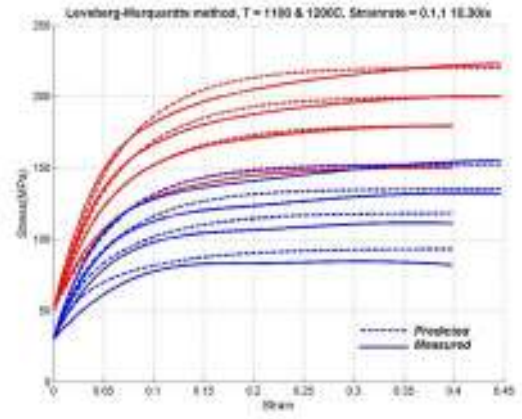


Fig. A4.8: Optimisation for Cr steel data at two different temperatures and four strain rates at each temperature

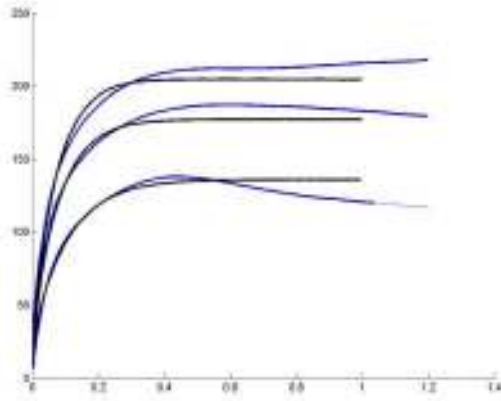


Fig. A4.9: Optimisation for C-Mn-Nb data at 850°C and at three strain rates

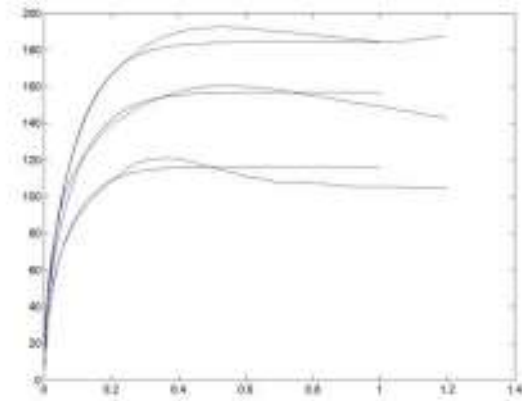


Fig. A4.10: Optimisation for C-Mn-Nb data at 900°C and at three strain rates

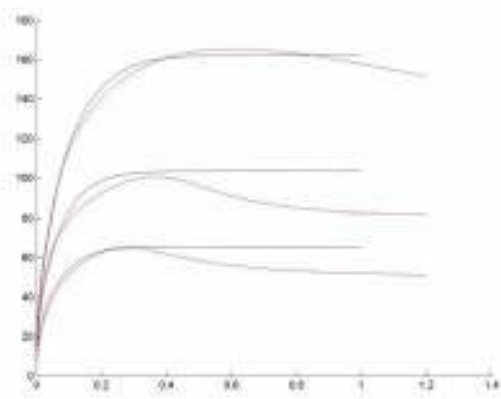


Fig. A4.11: Optimisation for C-Mn-Nb data at 1000°C and at three strain rates

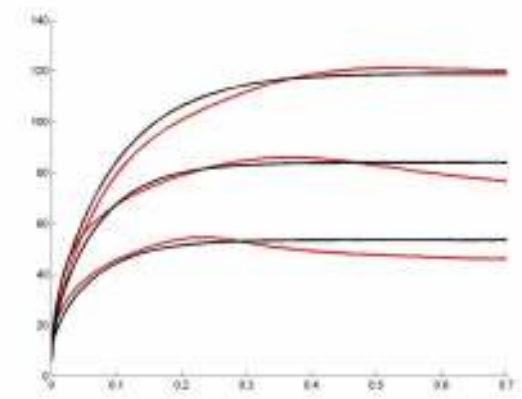


Fig. A4.12: Optimisation for C-Mn-Nb data at 1100°C and at three strain rates

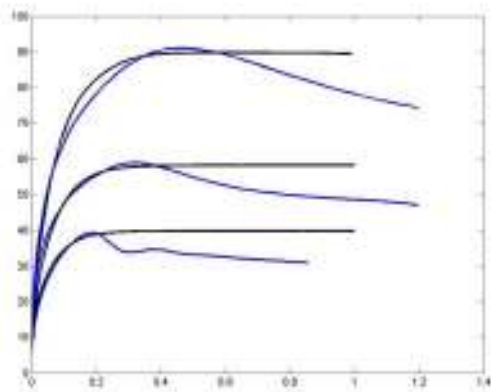


Fig. A4.13: Optimisation for C-Mn-Nb data at 1200°C and at three strain rates.

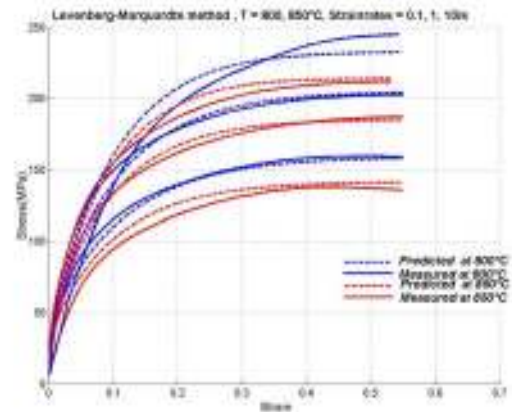


Fig. A4.14: Optimisation of two different temperatures and at three strain rates mean RMSE 5.79 MPa

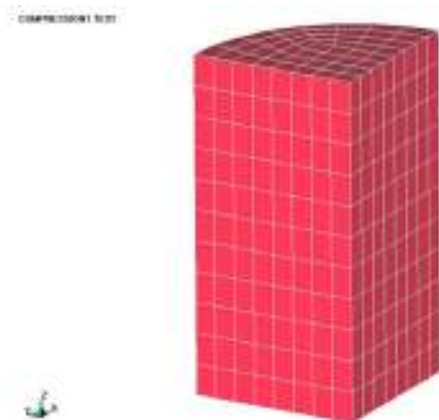


Fig. A4.15: FE model for simulating compression tests

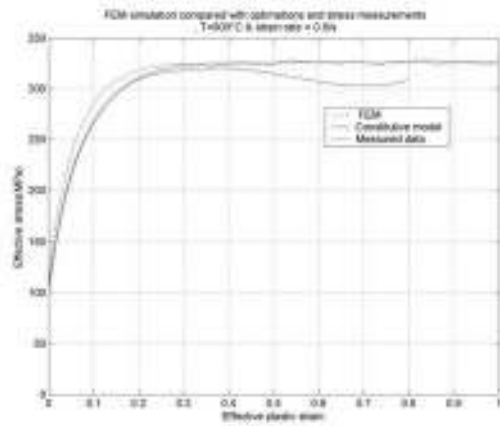


Fig. A4.16: Compression of Cr steel at 900°C and 0.8/s

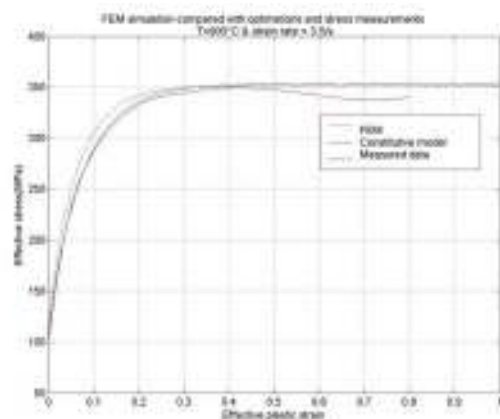


Fig. A4.17: Predictions at 900 and 3.5/s for Cr steel

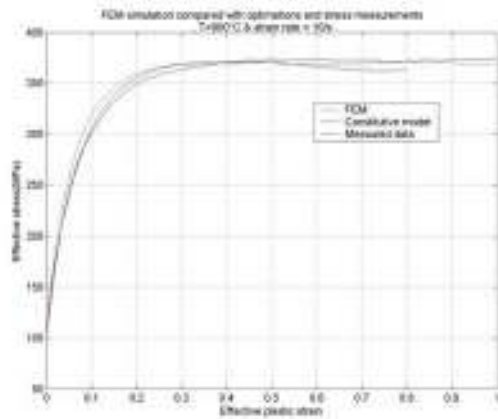


Fig. A4.18: Predictions at 900°C and 10/s for Cr steel

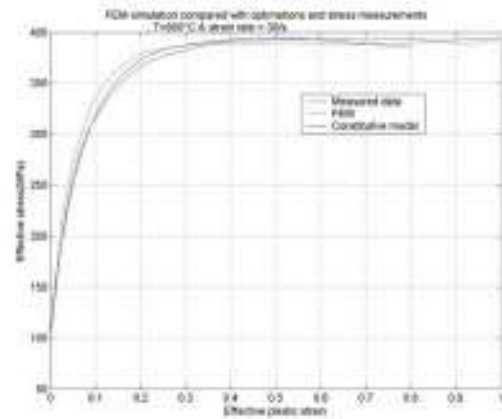


Fig. A4.19: Predictions at 900°C and 30/s for Cr steel

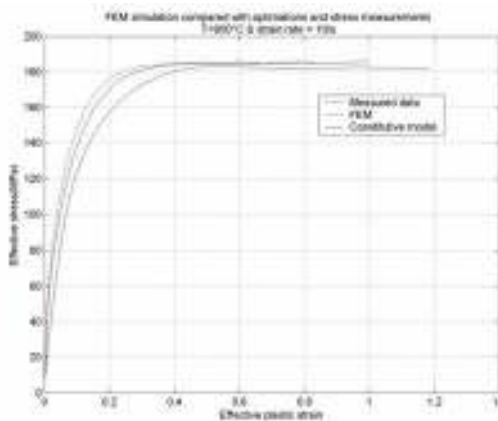


Fig. A4.20: Predictions at 900°C and at 10/s for C-Mn-Nb steel

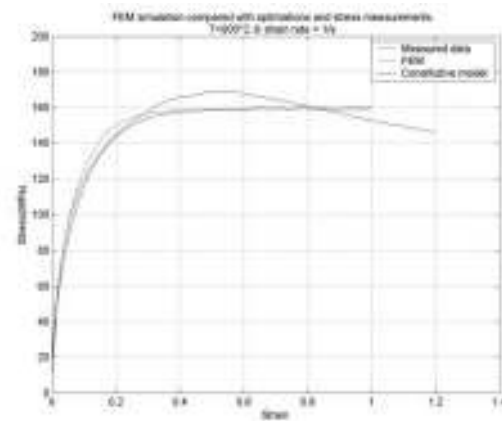


Fig. A4.21: Predictions at 900°C and at 1/s for C-Mn-Nb steel

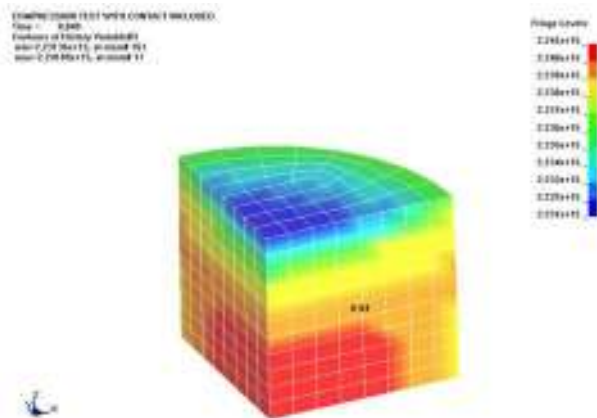


Fig. A4.22: Dislocation density at 900°C and at 10/s for C-Mn-Nb steel

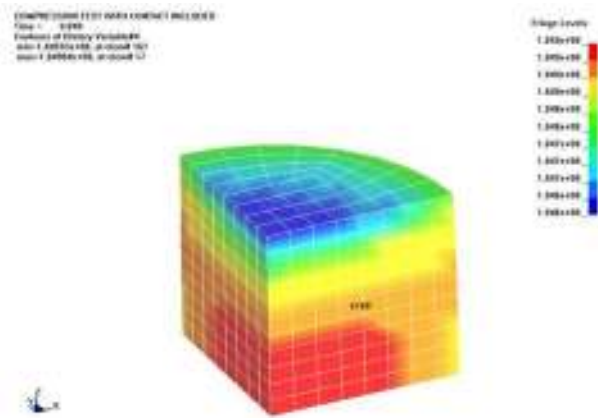


Fig. A4.23: Effective stress at 900°C and at 10/s for C-Mn-Nb steel

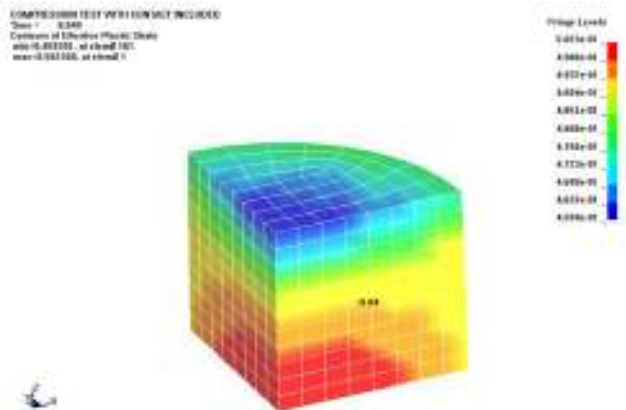


Fig. A4.24: Affective plastic strain at 900°C and at 10/s for C-Mn-Nb steel

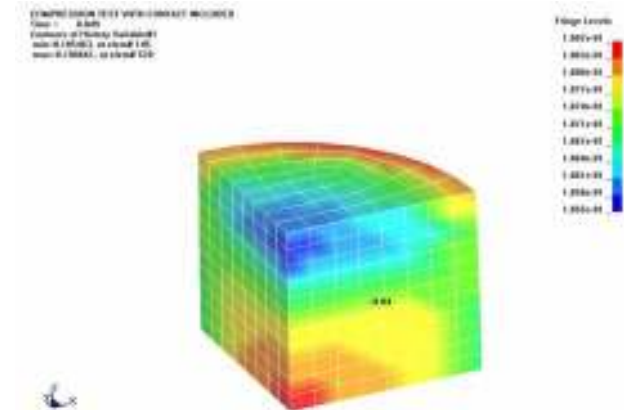


Fig. A4.25: Vacancy concentration at 900°C and at 10/s for C-Mn-Nb steel

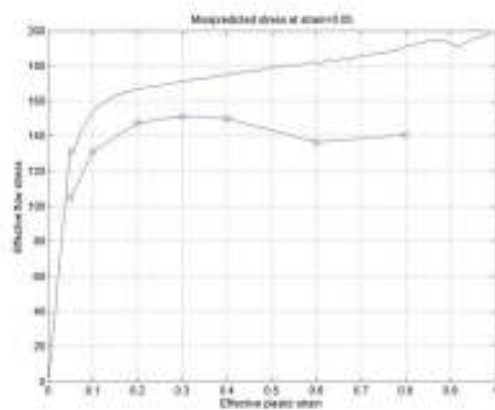


Fig. A4.26: Validation of simulations made with predicted initial frictional stress

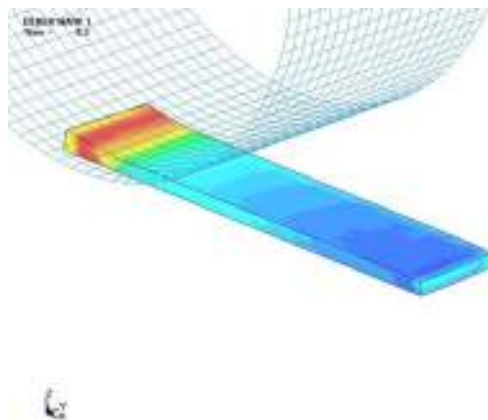


Fig. A4.27: benchmark testing case, colour bar indicates the dislocation density level

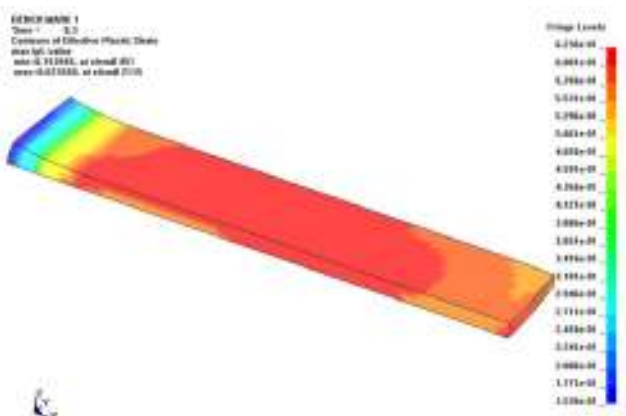


Fig. A4.28: Effective plastic strain for C-Mn-Nb steel

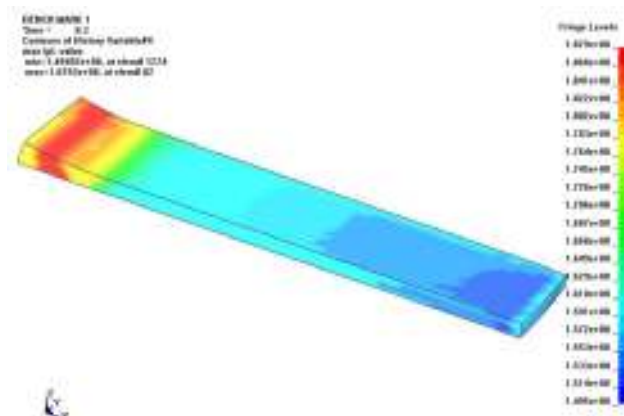


Fig. A4.29: Effective stress for C-Mn-Nb steel

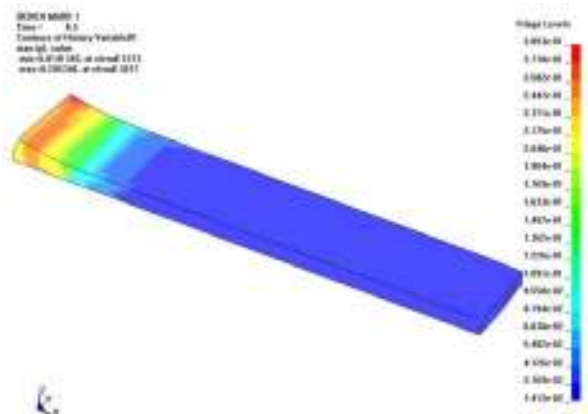


Fig. A4.30: Vacancy concentration for C-Mn-Nb steel

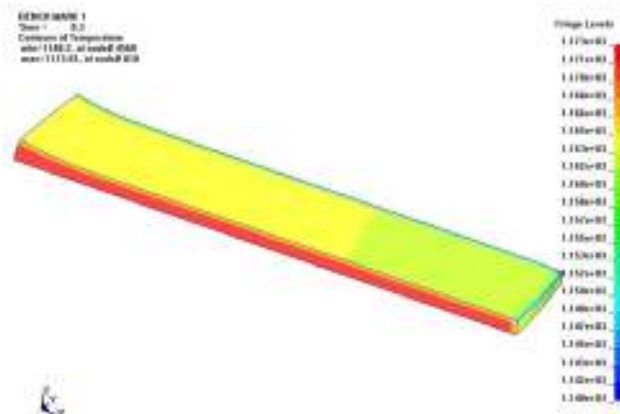


Fig. A4.31: Temperature for C-Mn-Nb steel without thermal conduction through the rolls

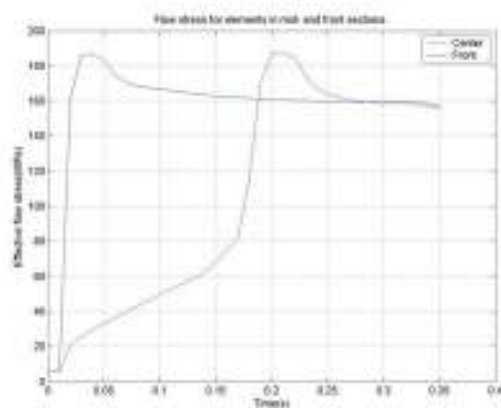


Fig. A4.32: Effective plastic stress in mid and front positions

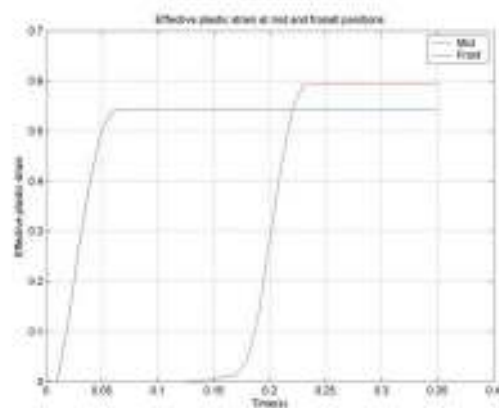


Fig. A4.33: Effective plastic strain in mid and front positions

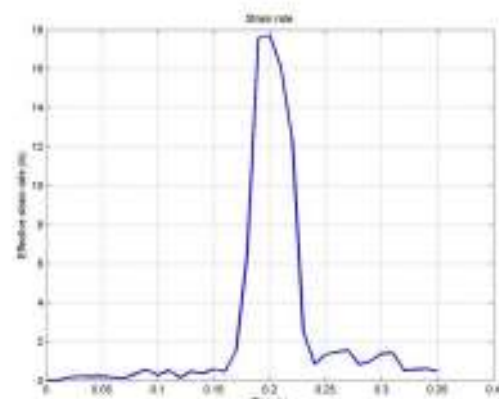


Fig. A4.34: Effective strain rate in mid positions

Appendix 5

Detailed Technical Report by TU BERGAKADEMIE Freiberg Authors: M Schmidtchen and R Kawalla

A5.1 Introduction

In the interpretation of industrial forming processes the coupling of single process steps plays an increasingly more important role. This applies particularly to continuous forming technologies, like the hot rolled strip and wire production or the open die forging. Since the influencing mechanical properties are increasingly carried out via the thermo mechanical treatment, the specific development of the microstructures gains a growing meaning.

The microstructure is for its part influenced by the tension and strain condition in the forming step as well as the interpass times, the material composition and the temperature regime.

The material transport during the forming passes is inhomogeneous over the cross section. A homogeneous initial state therefore results after the first passes in inhomogeneous strain distributions and therefore in an inhomogeneous change of the microstructure itself (locally different recrystallisation, grain growth and precipitation). These changes in microstructure influence the local material transport and the forming behaviour in the following passes now. In the common forming processes an additional stress and strain reversal will take place. Here a situation occurs which common simulation methods (empirical, numerical) can handle only insufficiently. The origin of this situation can be seen in the constitutive models used till now.

The aim of the partial projects for the individual partners in this ECSC multipartner project can be deduced from this point of view for (Fig. A5.1): A significant improvement on the simulation possibilities shall be found by improving the constitutive models.

Therefore rolling tests were carried out as benchmark for the simulation, material tests with different materials and test procedure as well as corresponding modellings in the individual partial projects. The material tests were carried out with different materials to guarantee a correct modelling of different softening behaviour. Therefore also different test procedures were used. The influence of different stress and strain states should be included here. By modelling of the results of the material tests and the rolling trials the quality of the received constitutive equations should be checked.

The main task of the TU BAF in this multipartner project consisted in delivering the experimental results from the rolling tests (geometry, rolling forces, torque and microstructure). An additional modelling of the strain hardening behaviour according to the Freiberg approach and its application in FE computations of single tests was the second topic of the partial program.

In detail the partial program of TU BA Freiberg was:

- **Experimental part:**

As part of the objectives set up for the benchmark exercise in this multi-partners ECSC project (Fig. A5.1), TU BERGAKADEMIE Freiberg has carried out experimental studies of the development of the austenitic grain size in high and low speed rod rolling mill with

involvement of other partners as well as low speed forward and reversal plate rolling (Fig. A5.2).

The experimental work of TU BERGAKADEMIE Freiberg is to be subdivided into three main phases:

1. Experimental work on rod rolling of medium C-Mn-steel with high and low rolling speed and on plate rolling of medium C-Mn-steel.
2. Investigation of the stress – strain relation of the medium C-Mn-steel used.
3. Visioplasticity investigations of the material flow in the roll gap during plate rolling.
4. Microstructural investigations of initial and final grain size.

- **Theoretical part:**

The experimental results obtained in the rolling tests will be used for a benchmark test of existing numerical models with strain reversal. In detail three steps were done:

- modelling of the stress – strain behaviour of C – Mn Steel by use of the Freiberg model
- Computation of the rolling trials done in the experimental part (empirical model for rod and plate rolling; FE - computation)
- Investigation of the softening behaviour at higher strain rate
- Computation with softening behaviour

A5.2 Experimental work

A5.2.1 Experimental equipment

A5.2.1.1 Continuous high speed rod rolling mill

For all trials the continuous high speed rolling mill at TU BA Freiberg in rod configuration was used (see Fig. A5.3 – Fig. A5.5). All data for high and low speed rolling are to be seen in [A5.1].

The roll pass design (round-oval; 10 mm ->6 mm) were held constant during all trials except trials at low speed using F1 and F2 at different strain. Geometry and rolling speed for all stands are presented in [A5.1] and Fig. A5.4.

At high speed (32 m/s) a configuration as to be seen in Fig. A5.3 was used (without water basin). At low speed (5 m/s) the final cooling was performed inside the turbulent cooling line. While the rolling speed was held constant (at 32 m/s or 5 m/s respectively) the heating temperature, heating time and entrance temperature at F1 was varied.

Temperature measurements were made during conductive heating, before entrance into stands F1 - F4 and at the laying head. In all stands the rolling force and rolling torque were measured.

Due to a fast increase of oxide scale and heat lost at the surface the measurement of surface temperatures was compared with data from thermocouple in the bulge at the same position. All measurements were done using the conductive heating of the continuous rolling mill.

The difference of the measured temperature increased. A maximum value of 140°C was obtained after heating for 10 min. This data were used to keep the temperature nearly constant during heating.

To optimise the cooling rate two series of tests were performed using a mixture of water and ice respectively water only. Water and ice was preferred due to thick layers of scale after long heating times. For short heating times (below 5 min) cold water cooling only is also useful.

Owing to 0.2% S a lot of segregation appeared. Therefore the austenitic grain size differs over the cross section. Examples of the obtained austenitic grain size are given in Fig. A5.12.

A5.2.1.2 Reversal rolling mill (two high)

For plate rolling trials a two high reversal rolling mill of TU BAF with roll diameter of 360 mm was used. In Fig. A5.6 the rolling technology is given. The specimens were reheated in a heating furnace. Before descaling and before entrance into the roll gap the surface temperature was measured. This was also done immediately after rolling. During rolling torque and rolling forces were measured. With the first temperature signal the time was measured. The interpass time could be deduced from the rolling force – time signal. After rolling the specimens were cooled immediately in a turbulent water bath.

A5.2.2 Continuous rod rolling – Effect of strain history; RX kinetik and rolling speed

For all trials the continuous high speed rolling mill at TU BA Freiberg in rod configuration was used (see Fig. A5.3 and Fig. A5.5). All data for high and low speed rolling are given in Table A5.2 –A5.6 of [A5.1].

The roll pass design (round-oval; 10 mm → 6 mm) was held constant during all trials except trials at low speed using F1 and F2 at different strain.

At high speed (32 m/s) a configuration as to be seen in Fig. A5.3 was used (without water basin). At low speed (5 m/s) the final cooling was performed inside the turbulent cooling line. While the rolling speed was held constant (at 32 m/s or 5 m/s respectively) the heating temperature, heating time and entrance temperature at F1 was varied.

Temperature measurements were made during conductive heating, before entrance into stands F1 - F4 and at the laying head. In all stands the rolling force and rolling torque were measured.

High speed rolling was performed using all stands F1-F4, water cooling and laying head. The reheating temperature was 1200°C and 1150°C respectively. The time for heating was also varied (approx. 10 s, 60 s; see also [A5.1] Table A5.3 – A5.6).

The length of the rod reheated was 5 m (insulating bricks/air). On both sides the cold parts outside the insulating bricks were cut. After cooling in air for:

- 1200°C -> 1100°C: ~10 s
- 1200°C -> 850°C: ~25 s
- 1150°C -> 1100°C: ~ 3 s
- 1150°C -> 850°C: ~25 s

the temperature for rolling was reached.

Rolling at **low speed** was performed using

- all stands F1 - F4 and water cooling,
- Stand F1 – F2 and water bath and
- Stand F1 only and water bath.

The length of the rod reheated between insulating bricks was 5 m. For rolling trials using all stands F1-F4 the reheated rod was completely rolled whereas during the trials with F1-F2 and F1 only the first part of the rod (approx. 1 m) was used.

Due to a low speed in the last stand F4 the rolled rods stopped inside the turbulent cooling line (trials using F1 – F4).

During all trials using F1-F2 or F1 only the rolled rod had to be pushed into stand F2 and into the water bath manually. Thus a break of about 2 s exists between each pass.

The reheating temperature was as similar as in high speed rolling, 1200°C and 1150°C respectively. Also the time for heating was varied (approx. 10 s, 60; see also Table A5.3 and A5.6. of [A5.1]).

The initial grain size determines significantly recrystallisation processes [A5.1]. Due to technical restrictions only reheating times of 10 s and max. 60 s at 1200°C could be realised. Thus a striven austenitic grain size of 200 µm wasn't reachable.

The microstructure of the steel grade used was inhomogeneous due to precipitates and segregation's. Thus, inhomogeneous recrystallisation will be expected.

For the given rolling speed and a fixed distance between the stands the interpass time is also given. To obtain different effects of the recrystallisation on the rolling forces the time for static and dynamic recrystallisation must be changed.

From Experimental data (in [A5.1]) obtained at Swinden Technology Centre (STC) the dependence of the recrystallisation time on temperature, initial grain size and strain is known.

The lower the rolling temperature and the initial grain size the lower the time for full recrystallisation. Increasing the strain will give the same results.

Whereas in [A5.1] an initial grain size of 100 μm was used to obtain $t_{0.95} \sim 0.006 \text{ s}$ (1100°C) and $t_{0.95} \sim 0.24 \text{ s}$ (850°C) the initial grain size for the rolling trials is in the range of 50 μm depending on the reheating parameter. Thus only for rolling trials at high rolling speed and low temperature un-recrystallised structures will be obtained. Variations in thermal conditions and rolling speed (interpass time) was used to affect the recrystallisation process during rolling (full and particular recrystallisation).

In the rolling trial two different sets for the rolling speed were used. The related sets of interpass time are given in Table A5.1. In [A5.1] the variations of the initial thermal conditions are reported.

During the rod rolling trials parameter were varied to study the effect of:

- Reheating temperature,
- Reheating time (at const. Temperature),
- Rolling speed,
- Temperature at the entrance of F1.

Strain variations were done due to the dependence of spreading on the rolling speed and temperature (const. roll gap geometry see [A5.1]).

From a plot of the rolling forces v time it is known that a stationary state was reached. Dependent on rolling speed and rolling temperature three principal trends of rolling forces were found (trend 1-3 in Fig. A5.7; A5.8). But all trials plotted in Fig. A5.7 results in similar final grain sizes (Table A5.2).

In general it is to be seen in the rolling force plots that the rolling speed and average temperature affects the recrystallisation process in a complex manner. But microstructural investigations of the material after the last pass (F4) had shown no significant differences in the final average austenitic grain size. Thus more detailed investigations of the microstructure after F1 and F2 are necessary to quantify the influence of the parameter given above.

The qualitative effect of reheating temperature and time can be derived from the development of the rolling forces during the whole rolling process (Fig. A5.9):

The trend curves in Fig. A5.9 represent the principal curves 1 (rolling speed 5 m/s) and 2 (rolling speed 30 m/s) in Fig. A5.7 and A5.8.

A higher reheating temperature results in lower rolling forces for constant rolling speed.

All trials plotted in Fig. A5.9 started at 850°C in the first pass F1. For different reheating temperatures similar cooling times (26 s ... 28 s) were necessary to reach 850°C. Due to different reheating temperatures grain growth reaches different stages, for specimen reheated up to 1200°C a more homogeneous microstructure but also more scale occur. For the free-cutting steel used the scale forces lubrication during rolling. So finally the reason for a lower rolling force can't be attributed only to grain growth. Therefore a detailed analysis of the microstructure obtained after the first two passes are necessary. Computation of a single pass and comparison with experimental results will also give some information about the significance of friction and recrystallisation.

During the variation of the rolling speed the ratio of the speed between the stands was held constant. The effect of speed variations is to be seen in Fig. A5.9 and A5.11.

At constant thermal initial conditions (Fig. A5.11) different trends in the rolling forces were obtained. Therefore only particular recrystallisation can occur. The trends in temperature (at F2-F4) for the trials given in Fig. A5.11 are similar but different in its absolute value. A possible reason is the shortening of the interpass time with increasing the rolling speed (Table A5.1).

At constant rolling speed the temperature at the entrance will also affect the trend of the rolling forces (Fig. A5.10). The rolling forces not only increase with higher temperatures. The trends differs also in the inclination in-between the passes. Similar results available at lower speed (see Fig. A5.7). These behaviour can be attributed to the material behaviour in the transformation range.

While differences in rolling force are visible, the final grain size (behind F4) will be only slightly affected by all parameter changes. Smaller grain sizes can be measured for higher reheating temperatures or higher reheating times and higher rolling speed at lower temperatures in the entrance of F1.

More detailed studies of the development of the microstructure and mechanical values especially of the first two passes and comparison with computational results will help to get more information.

Investigation of the grain growth behaviour

Due to a slow evolution of the grain size in CMnS steel grades during the reheating trials additional investigations of the grain growth behaviour were done. The final results are: no significant grain size evolution below 1200°C will occur. This is consistent with data from literature (Fig. A5. 13). CMnS steel grade tend to show abnormal grain size evolution.

For a detailed study dimensionless plots similar to the method given below for flat rolling is necessary. Therefore continuous and discontinuous stress strain curves must be predicted in upsetting tests.

A5.2.3 Pretests for plate rolling

Additional to the benchmark trials these pretest were performed:

Prerolling

The CMnS – steel grade delivered by TU Freiberg with an initial diameter of 25 mm was prerolled in an open three high train at TU Freiberg. Thus a 10 mm x 6 m rod was prerolled within similar parameters.

The CMnNb – steel grade delivered by Corus with plates of 30 mm thickness was particular prerolled down to 6.7 mm within two passes. Thus two sets of specimen with different initial thickness were available to investigate the significance of the rolling geometry, to vary the rolling parameter and to improve the cooling conditions after the last pass. The geometry of the rolling specimen was fixed by the geometry of the turbulent water bath. Prerolling was performed at similar strain – time regime at a rolling temperature above 900°C.

Prerolled rods and plates were cooled down in air.

Calibration of pyrometer

Temperature measurements were performed with pyrometer for rod rolling and plate rolling as well. During reheating or cooling parallel measurements with thermocouples were used for a calibration. Therefore the influence of oxide scale during temperature measurement could be reduced.

Investigation of phase transformation and cooling behaviour

To investigate the evolution of the austenitic grain size of CMnNb steel grades a fast cooling is necessary to obtain a martensitic microstructure. From a database of TU Freiberg the transformation behaviour of the CMnNb steel grade was computed (Fig. A5.16). As can be seen in Fig. A5.16 a cooling time of $t_{8/5} = 1.3$ s is necessary to obtain martensite. Lower cooling rates will result in a bainitic or ferritic/pearlitic microstructure. These microstructure will affect the investigation of the austenitic grain size.

On the basis of these data computations of a simplified cooling behaviour with maximum heat transfer were undertaken to predict the maximum thickness (exit of the roll gap after the last pass) for the cooling rate necessary ($t_{\max}=1.3$ s).

From Fig. A5.15 a cooling time $t_{8/5}$ valid for the center of the plate is to be seen in dependence on the thickness for a given width and length.

For a maximum cooling time of 1.3 s a maximum thickness of 15 mm was computed. Real heat transfer conditions will lower the cooling rate. Therefore a smaller maximum thickness must be used in the rolling trials. With these geometric conditions cooling trials with plates of 15 mm and 7 mm thickness were performed (Figs. A5.17 and A5.18). Finally we obtain from the cooling tests: only specimen with a final thickness below 7 mm will cool fast enough to obtain martensitic microstructure.

A5.2.4 Plate rolling – Effect of strain history; RX Kinetik and rolling parameter

Objective

The objective of the performed plate rolling trials was to investigate the connection between rolling parameter, RX kinetik and the evolution of the austenitic grain structure during forward or reversal rolling.

Main objective of the trials were to perform a benchmark test and to show the effect of different strain and stress histories on rolling forces and microstructural evolution.

Guidelines

From all partners the following guidelines for the plate rolling trials were specified:

Material:

CMnNb steel (0.15% C, 1.4% Mn, 0.37% Si, 0.03% Nb, 0.017% Cr, 0.002% Mo, 0.02% Ni, 0.033% Al, 0.001% Ti, 0.003% V, 0.002% Sn, 0.007% Cu, 0.006% N, 0.016% P, 0.009% S)

Plate size: W120 mm, L400 mm, H30 mm

Mill facility: Freiberg Reversing Stand Mill, STC Plate Mill

Rolling speed: 1 m/s; Roll diameter: 360 mm (Freiberg);

Rolling Schedule:

Heating temperature: 1200°C (30 min)

Rolling temperatures: 1000°C, 900°C, 850°C

Rolling trials:

- 1st trials: Only 1 pass: 40% reduction (12 mm/30 mm=40%), water quench.
- 2nd trials: 2 passes with the first pass 40% reduction (12 mm/30 mm=40%) and the second pass in the same direction for a further 50% reduction (9 mm/18 mm=50%), water quench.
- 3rd trials: 2 passes with the first pass 40% reduction (12 mm/30 mm=40%) and the second pass in the reversal direction for a further 50% reduction (9 mm/18 mm=50%), water quench.

Measured values:

- (a) Rolling force and torque
- (b) Temperature at entry and exit

Microstructure studies:

- (a) Grain size and microstructure photo BEFORE rolling (1000°C, 900°C, 850°C)
- (b) Grain size and microstructure photo AFTER rolling (water quench)

Plan and performance of the rolling trials

Rolling trials with two different initial thicknesses and rolling geometry were realized. In general as it was discussed above the specimen delivered from Corus (W120 mm, L400 mm, H30 mm) had to be cut into smaller pieces due to the geometry of the cooling bath and the maximum rolling force of the reverse rolling mill.

A length of 120 mm (30 mm initial thickness) and 162-180 mm (6.7 mm initial thickness) respectively were used. From Table A5.3 and A5.4 the initial width is given.

The different rolling conditions will result in different rolling parameter as discussed below. Focus was set on a similar ratio of contact length in rolling direction L_d and mean high of the roll gap h_m . The rolling schedule was set as given in the guidelines above. Also rolling conditions at lower strain were used:

- **Rolling trials using initial thickness of 30 mm**

In this series of trials the rolling schedule given in the guidelines using smaller specimen (H30 mm x W80 mm x L120 mm) was realised. Due to the final thickness of the rolling specimen of 15 mm and 8 mm respectively no martensitic microstructure was obtainable (Fig. A5.19). At an equivalent strain of $\varphi > 0.6$ per pass the rolling trials were performed in the region of dynamic recrystallisation. Decarbonisation could be observed in the surface region of the specimen (Fig. A5.19).

- **Rolling trials using optimised initial thickness of 6.7 mm**

This series of rolling trials were performed using prerolled specimen with a reduced initial thickness of 6.7 mm (H6.7 mm x W100 mm x L162...180 mm). After the 1 pass the specimen (thickness 5 mm) were cooled at a higher cooling rate. Thus the $t_{8/5}$ time in the center of the specimen is below 1.3 s and a martensitic microstructure is found.

Logarithmic strain

The logarithmic strain $\varphi = \ln \left[\frac{h_0}{h_1} \right]$ is used for computations. It is the total strain of a cross – section at the end of a pass and in its nature a mean value. In real processes the logarithmic strain is distributed inhomogeneously over the cross – section and the roll gap length. Thus for a calculation of the reference entities a mean value will be used.

Mean logarithmic strain rate

To calculate the mean strain rate in the roll gap an equation of Hoff and Dahl

$$\dot{\varphi}_m = \frac{v_w}{L_d} \varphi \quad \dots (A5.1)$$

was used. A good approximation within a range of $0 < \varphi < 0.5$ can be assumed [A5.6].

Interpass time

The interpass time is given from rolling force v time measurements. At the beginning of the 1st pass the count starts and ends at the beginning of the 2nd pass. To obtain the real interpass time the time for one rolling pass have to be subtracted from the counted time. In our case this are 0.013 ... 0.05 s and can be neglected.

Reheating temperature and time

Reheating was performed 30 s in air at 1200°C. A few seconds after reheating 1150°C were measured at the surface. This is due to scale growth and radiation. In the core of the specimen a temperature of 1200°C will last (A5.1).

Rolling temperature

The trials start at a higher temperature (rolling temperature + 50°C) before descaling. In front of the rolling mill the temperature at the entrance of the roll gap were measured. The pass starts when the temperature was exactly reached. At the exit of the roll gap temperature measurements were performed too.

Forward and reversal rolling

After the first pass the rolled specimen were manually moved back to the entrance of the roll gap and for reversal rolling turned around 180°. This manual operation allows the interpass time to be minimised compared to a change of the rolling direction of the mill. On the other hand deviations in the interpass time were caused. Due to additional steps the interpass time for reversal rolling is longer compared to that of forward rolling.

A5.2.5 Investigation and modelling of stress – strain curves in continuous upsetting tests

To plan the rolling trials, to compute typical parameter and to study the transformation behaviour near A_3 under forming conditions upsetting tests were performed. All specimen were reheated within 30 min at a constant temperature of 1200°C. Upsetting tests were done in a temperature range of 600°C up to 1200°C. To investigate the temperature range of phase transformation A_3 during the forming process plots of the yield stress at a constant equivalent strain for selected equivalent strain rates and temperatures were computed (Figs. A5.22 and A5.23).

The phase transformation was obtained at a temperature range of 750°C-800°C. To avoid phase transformation during rolling the trials must be performed at a temperature above 850°C.

The yield stress was calculated using the Freiberg model equation for the continuous stress – strain curves ($m_1 \dots m_8$ constants for approximation):

$$\sigma_F = A \cdot e^{m_1 \vartheta} \cdot \varphi^{m_2} \cdot e^{(m_4 / \varphi + m_7 \cdot \varphi)} \cdot (1 + \varphi)^{m_5 \vartheta} \cdot \dot{\varphi}^{m_8 \vartheta} \quad \dots \text{ (A5.2)}$$

The experimental data were fit to the model within a multiple approximation method (coefficients see Tables A5.5-A5.6).

A5.2.6 Softening behaviour

At TU Freiberg additional quasiadiabatic compression tests were performed to investigate the softening behaviour at higher strain rates. This value will be used to correct the dimensionless plot of the related forming resistance. Therefore compression tests at constant temperature (1000°C) were examined with different interpass times (0.5; 1; 5; 10; 50 s). The strain for the first step was 0.15 and 0.3 respectively. A strain rate of 1/s and 10/s was used.

The softening fraction was computed using the 0.2% offset method and temperature corrected stress – strain curves. Results for the $t_{0.5}$ parameter are to be seen in Figs. A5.31-A5.33.

At similar strain, smaller strain rate and higher temperature (1100°C; 1150°C) results from torsion test have shown higher values for the $t_{0.5}$ parameter [A5.3]. A similar dependence on strain and strain rate was predicted from torsion tests. Increasing prestrain strain rate and temperature will shorten the time for a certain value of fractional softening (Figs. A5.32 and A5.33). For increasing strain rate a fractional softening of 0.8 ... 1 was obtained for the interpass times of the plate rolling trials (5 ... 10 s).

A5.3 Results, discussion and simulation

A5.3.1 Comparison of rolling force and the microstructure for high speed rod rolling

The initial grain size determinates significantly recrystallisation processes [A5.1]. Due to technical restrictions only reheating times of 10 s and max. 60 s at 1200°C could be realised. Thus a striven austenitic grain size of 200 μm wasn't reachable.

The microstructure of the steel grade used was inhomogeneous due to precipitates and segregation's. Thus, inhomogeneous recrystallisation will be expected.

For the given rolling speed and a fixed distance between the stands the interpass time is also given. To obtain different effects of the recrystallisation on the rolling forces the time for static and dynamic recrystallisation must be changed.

From Experimental data (in [A5.1]) obtained at Swinden Technology Centre (STC) the dependence of the recrystallisation time on temperature, initial grain size and strain is known.

The lower the rolling temperature and the initial grain size the lower the time for full recrystallisation. Increasing the strain will give the same results.

Whereas in [A5.1] an initial grain size of 100 μm was used to obtain $t_{0.95} \sim 0.006$ s (1100°C) and $t_{0.95} \sim 0.24$ s (850°C) the initial grain size for the rolling trials is in the range of 50 μm depending on the reheating parameter. Thus only for rolling trials at high rolling speed and low temperature un-recrystallised structures will be obtained. Variations in thermal conditions and rolling speed (interpass time) was used to affect the recrystallisation process during rolling (full and particular recrystallisation).

In the rolling trial two different sets for the rolling speed were used. The related sets of interpass time are given in Table A5.1. In [A5.1] the variations of the initial thermal conditions are reported.

During the rod rolling trials parameter were varied to study the effect of:

- Reheating temperature,
- Reheating time (at const. Temperature),

- Rolling speed,
- Temperature at the entrance of F1.

Strain variations were done due to the dependence of spreading on the rolling speed and temperature (const. roll gap geometry see [A5.1]).

From a plot of the rolling forces v time it is known that a stationary state was reached. Dependent on rolling speed and rolling temperature three principal trends of rolling forces were found (trend1-3 in Figs. A5.7 and A5.8). But all trials plotted in Fig. A5.7 results in similar final grain sizes (Table A5.2).

A5.3.1.1 Reheating time and temperature

In general it is to be seen in the rolling force plots that the rolling speed and average temperature affects the recrystallisation process in a complex manner. But microstructural investigations of the material after the last pass (F4) had shown no significant differences in the final average austenitic grain size. Thus more detailed investigations of the microstructure after F1 and F2 are necessary to quantify the influence of the parameter given above.

The qualitative effect of reheating temperature and time can be derived from the development of the rolling forces during the whole rolling process:

The trend curves in Fig. A5.9 represent the principal curves 1 (rolling speed 5 m/s) and 2 (rolling speed 30 m/s) in Figs. A5.7 and A5.8.

A higher reheating temperature results in lower rolling forces for constant rolling speed.

All trials plotted in Fig. A5.9 started at 850°C in the first pass F1. For different reheating temperatures similar cooling times (26 s ... 28 s) were necessary to reach 850°C. Due to different reheating temperatures grain growth reaches different stages, for specimen reheated up to 1200°C a more homogeneous microstructure but also more scale occur. For the free-cutting steel used the scale forces lubrication during rolling. So finally the reason for a lower rolling force can't be attributed only to grain growth. Therefore a detailed analysis of the microstructure obtained after the first two passes are necessary. Computation of a single pass and comparison with experimental results will also give some information about the significance of friction and recrystallisation.

A5.3.1.2 Rolling speed

During the variation of the rolling speed the ration of the speed between the stands was held constant. The effect of speed variations is to be seen in Figs. A5.9 and A5.11.

At constant thermal initial conditions (Fig. A5.11) different trends in the rolling forces were obtained. Therefore only particular recrystallisation can occur. The trends in temperature (at F2-F4) for the trials given in Fig. A5.11 are similar but different in its absolute value. A possible reason is the shortening of the interpass time with increasing the rolling speed (Table A5.1).

A5.3.1.3 Temperature at the entrance of F1

At constant rolling speed the temperature at the entrance will also affect the trend of the rolling forces (Fig. A5.10). The rolling forces not only increase with higher temperatures. The trends differs also in the inclination in-between the passes. Similar results available at lower speed (see Fig. A5.7). This behaviour can be attributed to the material behaviour in the transformation range.

A5.3.1.4 Conclusions

While differences in rolling force are visible, the final grain size (behind F4) will be only slightly affected by all parameter changes. Smaller grain sizes can be measured for higher reheating temperatures or higher reheating times and higher rolling speed at lower temperatures in the entrance of F1.

A5.3.2 Plate rolling trials

Rolling forces

For the final evaluation mean values of the rolling forces were used. Due to a short maximum length of the specimen compared to the contact length small regions of a stationary state were observed in the first pass.

Evaluation method

In Fig. A5.19 rolling forces v rolling temperatures are plotted. Only an increase of the rolling forces with decreasing temperatures is to be seen. Other information couldn't derived from this kind of plots. For comparative studies dimensionless plots will be derived to exclude the influence of deviations in roll geometry and temperature. After [A5.6] and [A5.5] the ratio of the mean deformation resistance k_{Wm} and mean yield stress k_{Fm} is plotted in dependence on the ratio of the contact length L_d and the mean thickness of the roll gap h_m (Fig. A5.24).

$$\frac{k_{Wm}}{k_{Fm}} \left(\frac{L_d}{h_m} \right) : \quad \frac{L_d}{h_m} = \frac{\sqrt{R(h_0 - h_1)}}{(h_0 + h_1)/2} \quad \dots (A5.3)$$

To derive the basic ideas the analogy of upsetting test and rolling can be used: The pressure distribution and thus the force depend on the ratio of contact length and thickness of the specimen, the friction stress and the mean yield stress.

Under an aspect of similarity the rolling force F is a function of the rolling geometry, rolling speed, frictional forces and yield stress only. The effect of temperature and strain are included in these parameters. Thus we will set:

$$F = f(R, B, h_0, h_1, v_w, \mu(\vartheta, \text{material composition}, \dots), k_{Fm}(\text{material composition}, \varphi, \dot{\varphi}, \vartheta))$$

Using the Pi – theorem ([A5.7]) in dimensionless form we are able to reduce the number of independent entities by 3. To rewrite this relation for constant rolling speed only 4 dimensionless entities are necessary:

$$\frac{k_{Wm}}{k_{Fm}} = \frac{F}{k_{Fm}(\text{material composition}, \varphi, \dot{\varphi}, \vartheta) B L_d} = f\left(\frac{L_d}{h_m}, \frac{B}{h_0}, \mu\right) \quad \dots (A5.4)$$

resp.
$$\frac{k_{Wm}}{k_{Fm}} = \frac{F}{k_{Fm}(\text{material composition}, \varphi, \dot{\varphi}, \vartheta) B L_d} = f\left(\frac{R}{h_1}, \frac{B}{h_0}, \mu\right)$$

Figure A5.25 shows a typical plot for the dependence of the normal pressure und thus the change of the rolling force on the aspect ratio L_d/h_m .

For comparative studies the mean forming resistance k_{Wm} is given from experimental data by

$$k_{Wm} = \frac{F}{B L_d} \quad \dots (A5.5)$$

with the width of rolling specimen B . The mean value of the yield stress k_{Fm} will be computed for each pass by

$$k_{Fm} = \frac{1}{\varphi(L_d) - \varphi_0} \int_{\varphi_0}^{\varphi(L_d)} k_F(\varphi, \dot{\varphi}, \vartheta) d\varphi \quad \dots (A5.6)$$

In Fig. A5.24 the relation of the ratio k_{Wm}/k_{Fm} to the aspect ratio L_d/h_m is given.

Two main parts of interest are to be seen. An interpretation is easy derived from elementary theory of plasticity:

1. $L_d/h_m \leq 1$: In this range the forming resistance is mainly affected by geometric conditions and the yield stress. Deviations in the stress-strain behaviour but in geometrical conditions also will result in significant deviations in k_{Wm}/k_{Fm} .
2. $L_d/h_m > 1$: In this range of L_d/h_m the ratio k_{Wm}/k_{Fm} depends roughly linear on L_d/h_m . From elementary theory of plasticity it is known friction forces are dominant in this parameter range. The trials were performed in this region. Therefore during the trials attention was taken on a complete and homogeneous descaling process after reheating.

In general for plate rolling without tensional stresses at the entrance or exit of the roll gap the ratio k_{Wm}/k_{Fm} will depend on the following parameter [A5.5]:

$$\frac{k_{wm}}{k_{Fm}} \uparrow \Leftrightarrow k_{Fm} \downarrow, v_w \uparrow, \mu \uparrow, \frac{L_d}{h_m} > 1 \uparrow, \frac{R}{h_1} \uparrow, \frac{B}{h_0} \uparrow \quad \dots (A5.7)$$

The length scales are given above, whereas the frictional coefficient μ will increase in the hot forming region with increasing normal pressure, decreasing rolling temperature and strain rate. A stress strain behaviour and thus the mean yield stress of the pass will depend on:

$$k_{Fm} \downarrow \Leftrightarrow \varphi_0 \downarrow, \varphi(L_d) - \varphi_0 \downarrow, \dot{\varphi} \downarrow, \vartheta \uparrow, \text{initial grain size} \uparrow \quad \dots (A5.8)$$

On the basis of these discussions the results of the rolling trials will be analysed below:

1st pass

In Fig. A5.26 the data of the 1 and 2 pass are plotted in dimensionless form. Additional parts of the upper and lower curve from Fig. A5.24 are inserted. The data of the 1 passes of both rolling series (30 mm and 7 mm initial thickness) are located near the lower curve.

These results will prove again the basic assumptions of the similarity model.

Observed deviations in the 1 pass will result in deviations of the computation of the mean width, different cooling conditions and thus a change of the frictional stresses and mean yield stress. Due to a temperature gradient the specimen will be at a higher temperature in the core compared to surface. Thus the computed mean yield stress will be higher. The same effect will result from the approximation of the stress-strain curves. At a strain of 0.6 and more the model will give higher values than the experimental data (+5...20 MPa) at the strain rate and temperatures used. With increasing temperature the frictional stresses will also decrease. These effects will result in a lower ratio k_{Wm}/k_{Fm} .

2nd pass

In general the results in the 2 pass are located at the upper boundary or higher in Fig. A5.28. Data obtained in the 1 trials with initial thickness of 30 mm will give higher values for k_{Wm}/k_{Fm} than data from the 2 series. Furthermore differences in the ratio of k_{Wm1}/k_{Fm1} and k_{Wm2}/k_{Fm2} of the 1 and 2 pass were predicted (Fig. A5.27).

For a similar aspect ratio $L_d/h_m \approx 3$ in the second pass the dimensionless ratio k_{Wm2}/k_{Fm2} of the trials using 30 mm initial thickness shows higher values than for trials with 7 mm initial thickness. Both series differ in strain history and aspect ratio B/h_0 (Table A5.3 and A5.4, Figs. A5.26 and A5.27).

The mean yield stress was calculated using the Freiberg model equation for the continuous stress –strain curves were the softening behaviour due to interpass time was neglected. In the future work discontinuous upsetting test will follow.

To identify the role of strain history and recrystallisation we have to discuss the effect of geometry and friction first.

From (7) the ratio k_{Wm}/k_{Fm} will increase for a constant aspect ratio L_d/h_m for decreasing temperature, aspect ratio B/h_0 and increasing ratio of real to computed stress – strain – behaviour:

- A decrease in rolling temperature will result in increasing frictional stresses. This will result an increasing ratio k_{Wm}/k_{Fm} . Different interpass times result in different frictional conditions due to the growth of oxide scale of different thickness.
- A decreasing aspect ratio B/h_0 for the different series and thus in the stress state (2D or 3D) have to result in a reduced ratio k_{Wm}/k_{Fm} . From experimental data at constant $L_d/h_m \approx 3$ we obtain a contrary behaviour (Fig. A5. 26). Thus comparing both rolling series the aspect ratio B/h_0 is not the main parameter. Both series differ in the strain history (Fig. A5.27). In the 1 series with initial thickness 30 mm significant increase in the ration k_{Wm}/k_{Fm} is obtained. This is on the one hand due to a rise in L_d/h_m but also due to a significant change in the stress state ($B/h_0 \approx 3 \Rightarrow 5.5$). In the 2 series with initial thickness 7 mm in the

1 pass a two dimensional stress state is reached ($B/h_0 \approx 15$). Thus a further increase in B/h_0 will not significantly affect the stress state. No increase of the width was measured for $B/h_0 > 15$ but also for one trial in the 1 series with $B/h_0 \approx 5.4$ and a strain of 0.35 in the second pass similar to the second series. This single trial gives a higher k_{wm}/k_{Fm} in the second pass too. This will be an indication of the influence of friction and stress-strain-behaviour.

- Changes in the real stress - strain – behaviour will affect the mean yield stress and thus the value of k_{wm}/k_{Fm} because this behaviour was assumed unchanged for the calculation of k_{Fm} . For these changes a few reasons are possible:
 - (i) Deviation between measured and real surface temperature and between surface and core as well.
 - (ii) Softening processes:
 - For a 100% removal of work hardening at an interpass time of more than 5 s like in the rolling trials will be enough at 1000°C. Below 950°C an inhibition of static softening due to forming induced precipitations will occur.
 - Grain refining due to recrystallisation in combination with static softening in the 1 pass will result in a steeper increase of the stress strain curve in the 2 pass compared to curves for a larger grain size. This results in a higher real k_{wm} in the 2 pass. Whereas the 1 series were performed in the region of dynamic recrystallisation ($\varphi = 0.6$ in the 1. pass, Fig. A5.27) the trials of the 2 series were performed without full DRX. Therefore higher values of k_{wm}/k_{Fm} in the 1 series are possible.
 - Grain growth e.g. at 1000°C after a full RX is a process working in the opposite direction of grain refining and lowers the ratio of k_{wm}/k_{Fm} . Higher strain rates and strain will force grain growth at constant interpass time. The stress – strain curve of such trials will be close to a continuous curve. Particularly this effect will be possible for the first series with initial thickness of 30 mm at 1000°C (Fig. A5.27).
 - An inhomogeneous distribution of real strain and strain rate and strain history may also affect the ratio k_{wm}/k_{Fm} .

Microstructural investigation:

In the plate rolling trials several specimens of similar geometry and similar initial microstructure were reheated at the same temperature and same time. Then the specimens were rolled at the same rolling geometry in the first and second pass. The difference between both groups of trials was the rolling direction in the second pass. Therefore different microstructures were predicted.

From microstructural investigations of the rolled specimens we obtain a smaller grain size after reversal rolling than for forward rolling (Figs. A5.28 and A5.29). This behaviour is

independent of the rolling temperature. An inhomogeneous grain size distribution was obtained. The general differences could be due to local differences in the strain pass history.

A5.3.3 Computational studies and online Visio plasticity

In several FE computations using the FE code MARC including subroutines for the stress-strain behaviour the rolling trials described in A5.2 were analysed (Figs. A5.34 and A5.35). Due to differences in the shear deformation during the second pass deviations between forward and reversal rolling were obtained. Furthermore the flow lines during the passes were computed. These results were used for comparison with results obtained in Visio plasticity measurements.

The method of visioplasticity was optimised in several rolling trials for online measurements in hot rolling. The aim was to predict the real flow behaviour in dependence of the rolling direction and to exclude frictional behaviour.

From computational studies the origin of the inhomogeneity and the differences in the RX behaviour was identified.

The experimental results were used for comparison with FE - simulations. In Fig. A5.36 – Fig. A5.38 results obtained in rolling trials and 2D – FE simulations are shown. Isotropic hardening and von Mises material were used in FE – computations. The deformation of the flow lines and the wire grid show similar results. Small deviations are to be seen near the exit of the rolling gap in the middle of the rolling sample as well as in an area of $\frac{1}{4}$ of the sample thickness below the surface.

The origin of the deviations is the out - of - plane displacement of the specimen surface (Fig. A5.38(b)). Other reasons are frictional effects and the influence of thermal boundary conditions. Further simulations using 3D FE –Models will help to overcome these deviations.

A5.4 Citations

A5.1	Technical report No.1 of the ECSC Project 7210.PR/291 (D3.05/01); "Constitutive modelling complex loading in metal forming processes", March 2002.
A5.2	Technical report No. 2 of the ECSC Project 7210.PR/291 (D3.05/01); "Constitutive modelling complex loading in metal forming processes", August 2002.
A5.3	Technical report No. 3 of the ECSC Project 7210.PR/291 (D3.05/01); "Constitutive modelling complex loading in metal forming processes" , March 2003.

- A5.4 Gutierrez, I, Lopez, B and Rodriguez-Ibabe, J M: "Grain growth phenomena during steel processing", in: Proc. 1. Joint International Conf. Recrystallisation and Grain Growth, Gottstein, G; Molodov, D.A., 2001, pp145 – 154.
- A5.5 Hensel, A (ed.): „Technologie der Metallformung – Eisen- und Nichteisenwerkstoffe“, VEB Deutscher Verlag für die Grundstoffindustrie Leipzig, 1990.
- A5.6 Hensel, A and Spittel, T: „Kraft- und Arbeitsbedarf bildsamer Formgebungsverfahren“, VEB Deutscher Verlag für die Grundstoffindustrie Leipzig, 1978.
- A5.7 Spurk, J H: „Dimensionsanalyse in der Strömungsmechanik“, Springer Verlag, 1992.

Table A5.1: Interpass time in continuous rod rolling

v_{F4}	F1-F2	F2-F3	F3-F4	F4-Cooling
5.39 m/s	0.6 s	1.1 s	0.26 s	1.17 s
32.36 m/s	0.1 s	0.2 s	0.04 s	0.19 s

Table A5.2: Average austenitic grain size of rod rolling trials

No.	Average austenitic grain size	Standard deviation	skewness
1a	9.7 μm	--	--
1b	11.44 μm	5.51 μm	1.02 μm
3	11.16 μm	5.48 μm	0.86 μm
4	8.1 μm	3.05 μm	0.76 μm
5	11.67 μm	5.27 μm	0.98 μm
6	9.45 μm	4.27 μm	1.17 μm
10	13.1 μm	6.31 μm	0.82 μm
11	11.03 μm	5.18 μm	1.12 μm

Table A5.3: Rolling conditions for plate rolling (1 Series)

Rolling speed 1 m/s; diameter of roll 360 mm

Initial thickness h_0 [mm]	Final thickness h_1 [mm]	B_m [mm]	k_{Fm} [MPa]	L_d/h_m [-]	k_{wm}/k_{Fm} [-]	$\dot{\phi}_{m1}$ [1/s]	rolling temp. In [°C]	rolling temp. Out [°C]	B_m [mm]	k_{Fm} [MPa]	L_d/h_m [-]	k_{wm}/k_{Fm} [-]	ϕ_2 [-]	$\dot{\phi}_{m2}$ [1/s]	interpass time [s]	reversal
29.5	15.9	82.95	159	2.2088	1.37389	0.6244	12.678	1005	938							1
29.5	8.5	85.75	164	2.2088	1.3555	0.6244	12.678	994	923	162	2.9729	2.37205	0.61994378	17.028	8.76	2
29.5	8.3	86.1	162	2.2088	1.37096	0.6244	12.678	1000	928	171	3.0392	2.22615	0.64375443	17.449	6.2	2r
29.5	7.6	92.55	158	2.2088	1.39955	0.6244	12.678	1000	950	160	3.2759	2.31189	0.73186169	18.987	4.78	2
29.5	7.2	87.95	157	2.2088	1.43277	0.6244	12.678	1004	948	158	3.4148	2.4932	0.78592891	19.918	5.28	2
29.5	7.4	85.25	154	2.2088	1.49626	0.6244	12.678	1003	966	161	3.345	2.47338	0.75852994	19.447	4.94	2
29.5	7.3	87.85	170	2.2088	1.35021	0.6244	12.678	998	887	164	3.3798	2.51916	0.77213559	19.681	7.6	2r
29.5	7.2	89	167	2.2088	1.31355	0.6244	12.678	1002	900	156	3.4148	2.58759	0.78592891	19.918	7.2	2r
29.5	7.3	87.65	168	2.2088	1.32721	0.6244	12.678	1000	898	154	3.3798	2.54955	0.77213559	19.681	6.58	2r
29.5	7.2	87	193	2.2088	1.37522	0.6244	12.678	923	861	191	3.4148	2.4625	0.78592891	19.918	5.24	2
29.5	7.6	86.6	198	2.2088	1.43786	0.6244	12.678	898	868	212	3.2759	2.29766	0.73186169	18.987	5.7	2
29.5	7.3	84.6	190	2.2088	1.4191	0.6244	12.678	913	885	197	3.3798	2.37093	0.77213559	19.681	4.8	2
29.5	7.6	83.45	199	2.2088	1.45688	0.6244	12.678	895	866	207	3.2759	2.33568	0.73186169	18.987	4.24	2
29.5	7.5	85.5	210	2.2088	1.34523	0.6244	12.678	899	814	214	3.3103	2.28847	0.74510692	19.216	6.48	2r

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29.5	7.9	88	215	2.2088	1.30205	0.6244	12.678	901	797	88	206	3.1735	2.30213	0.69314718	18,315	5,28	2r
29.5	7.8	87.4	241	2.2088	1.3437	0.6244	12.678	844	760	87.4	250	3.2075	2.24673	0.70588621	18,537	6,6	2
29.5	11.1	83.9	224	2.2088	1.41853	0.6244	12.678	848	817	83.9	250	2.1432	1.64516	0.35306483	12,024	5,3	2
29.5	7.7	86.5	227	2.2088	1.40262	0.6244	12.678	844	809	86.5	248	3.2416	2.21372	0.71878961	18,761	8,56	2
29.5	8	86.85	244	2.2088	1.27788	0.6244	12.678	849	745	86.85	241	3.1397	2.21635	0.6805684	18,095	7,02	2r
29.5	7.8	86.8	238	2.2088	1.28272	0.6244	12.678	851	763	86.8	241	3.2075	2.26293	0.70588621	18,537	7,54	2r
29.5	7.7	85.45	241	2.2088	1.23809	0.6244	12.678	847	757	85.45	243	3.2416	2.13067	0.71878961	18,761	6,46	2r
29.5	16	82.05	242	2.1668	1.32304	0.6118	12.411	847	750								1
29.5	16	83.55	232	2.1668	1.36052	0.6118	12.411	842	794								1
29.5	16	82.75	240	2.1668	1.32687	0.6118	12.411	845	763								1
29.5	15.9	85.6	222	2.1796	1.31245	0.6181	12.492	894	778								1
29.5	16	84.8	218	2.1668	1.35523	0.6118	12.411	894	792								1
29.5	15.9	84.4	218	2.1796	1.35883	0.6181	12.492	895	792								1
29.5	15.5	84.6	171	2.2311	1.34395	0.6436	12.82	1000	885								1
29.5	15.6	83.5	174	2.2182	1.34298	0.6371	12.737	994	877								1
29.5	15.5	83.3	170	2.2311	1.34341	0.6436	12.82	1000	891								1
29.5	15.6	85.05	167	2.2182	1.38222	0.6371	12.737	1004	897								1
29.5	14.4	85.95	170	2.3751	1.33244	0.6244	13.756	998	891								1
29.5	7.5	86.85	170	2.2088	1.33861	0.6244	21.763	1002	888	86.85	154	3.3103	2.45228	0.74510692	19,216	5,92	2

Table A5.4: Rolling conditions for plate rolling (2 Series)

Rolling speed 1 m/s; diameter of roll 380 mm

Initial thickness h_0 [mm]	Final thickness h_1 [mm]	B_m [mm]	k_Fm [MPa]	L_d/h_m [-]	k_wm/k_Fm	ϕ_1 [-]	$\dot{\phi}_{m1}$ [1/s]	rolling temp. In [°C]	rolling temp Out [°C]	B_m [mm]	k_Fm [MPa]	L_d/h_m [-]	k_wm/k_Fm	ϕ_2 [-]	$\dot{\phi}_{m2}$ [1/s]	interpass time [s]	reversal
6.7	5	101.25	157.47	3.0722	1.7309	0.2927	16.285	990	937	101.25							1
6.74	4.98	102.55	158.1	3.1206	1.7202	0.3026	16.549	990	940	102.55							1
6.68	4.88	99.2	164.63	3.1995	1.6423	0.314	16.978	987	912	99.2							1
6.55	3.75	100.33	153.96	2.9558	2.0837	0.274	15.649	990	944	100.33	214	3.5355	2.3092	0.28367405	18.739	4.58	2
6.8	3.96	100.53	163.71	3.1197	2.0182	0.3115	16.546	981	920	100.53	212	3.1527	2.017	0.22918587	16.67	3.06	2
6.85	3.89	99.45	166.19	3.1497	1.8622	0.3188	16.712	987	903	99.45	224	3.2814	2.1358	0.24702073	17.363	5.26	2r
6.86	3.85	101.33	167.73	3.1556	1.7715	0.3203	16.744	983	899	101.33	228	3.3544	1.9481	0.25735674	17.757	3.38	2
6.84	3.78	99.375	158.77	3.1437	1.9186	0.3174	16.679	1013	920	99.375	225	3.4814	1.8204	0.27570588	18.445	5.1	2r
6.91	3.85	100.53	155.57	3.1846	1.9333	0.3275	16.905	1011	947	100.53	222	3.3544	1.9202	0.25735674	17.757	6.24	2r
6.96	5.11	98.775	196.42	3.1066	1.8392	0.309	16.48	900	823	98.775							1
6.67	4.94	100.63	190.16	3.1232	1.7958	0.3003	16.561	900	850	100.63							1
6.73	5.03	99.3	188.62	3.0565	1.8359	0.2912	16.2	900		99.3							1
6.77	3.97	101.7	194.13	3.1013	1.7049	0.3071	16.445	900	833	101.7	260	3.1342	2.1089	0.2266638	16.571	4.96	2
6.82	3.97	99.925	193.14	3.1318	1.8708	0.3144	16.613	903	840	99.925	270	3.1342	1.8483	0.2266638	16.571	5.7	2
6.84	4.02	96.95	199.91	3.1437	1.8804	0.3174	16.679	891	821	96.95	280	3.0412	2.0248	0.21414799	16.071	6.12	2

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6.8	3.95	95.975	196.58	3.1197	1.7604	0.3115	16.546	900	824	95.975	283	3.1712	2.0363	0.23171431	16.769	7.6	2r
6.69	3.98	101.03	191.79	3.0505	1.81	0.2952	16.167	893	843	101.03	259	3.1157	2.2752	0.22414807	16.471	6.18	2r
6.74	4.02	99.475	191.68	3.0825	1.879	0.3026	16.342	895	847	99.475	266	3.0412	2.0746	0.21414799	16.071	5.22	2r
6.72	5.09	98.275	209.08	2.9802	2.0105	0.2778	15.786	850	788	98.275							1
6.73	5.01	97.525	216.86	3.0797	1.9564	0.2951	16.326	841	777	97.525							1
6.68	5.14	97.175	205.05	2.8943	2.1007	0.2621	15.32	850	792	97.175							1
6.72	4.06	96.125	211.02	3.0698	1.9436	0.2997	16.272	850	797	96.125	297	2.9662	2.2022	0.20424692	15.669	7.58	2
6.77	4.09	102.2	210.5	3.1013	1.8267	0.3071	16.445	850	805	102.2	315	2.9095	2.1416	0.19688492	15.366	6.4	2
6.77	4.12	100.85	218.29	3.1013	1.8994	0.3071	16.445	850	770	100.85	305	2.8525	2.0491	0.18957673	15.061	4.48	2
6.77	4.11	101.13	214.7	3.1013	1.7646	0.3071	16.445	850	786	101.13	295	2.8715	2.0985	0.19200686	15.163	5.28	2r
6.76	4.18	99.775	212.01	3.095	1.7774	0.3056	16.411	850	797	99.775	303	2.737	2.0982	0.17511864	14.445	5.46	2r
6.68	4.09	100.43	211.5	3.044	1.7883	0.2937	16.131	850	790	100.43	302	2.9095	2.0068	0.19688492	15.366	5.26	2r
6.7	4.95	98.925	198.32	3.1304	1.817	0.3027	16.601	895	816	98.925							1

Table A5.5: Calculation constants for stress – strain curves

material	CMnNb steel grad
Constants of thermodynamic factors	
<u>Ansatz 8</u>	
constant A	7917.86
temperature K ϑ :	
m_1	-0.00555086
strain K φ :	
m_2	-0.049047
m_4	-0.00569492
m_5	0.00313907
m_7	-1.21795
Strain rate K φ'	
m_8	0.000149325
multiple correlation	0.971030
Range of validity	
Forming conditions	
ϑ [°C]	600...770
φ	0,05 ... 1,2
$\dot{\varphi}$ [s ⁻¹]	0,1 ... 500

Table A5.6: Calculation constants for stress – strain curves

material	CMnNb steel grad
Constants of thermodynamic factors	
<u>Ansatz 8</u>	
constant A	2638.35
temperature K ϑ :	
m_1	-0.00198524
strain K φ :	
m_2	0.516978
m_4	0.00107422
m_5	-0.00279135
m_7	0.553024
Strain rate K φ'	
m_8	0.000137173
multiple correlation	0.987611
Range of validity	
Forming conditions	
ϑ [°C]	770 ... 1200
φ	0,05 ... 1,0
φ' [s ⁻¹]	0,1 ... 500

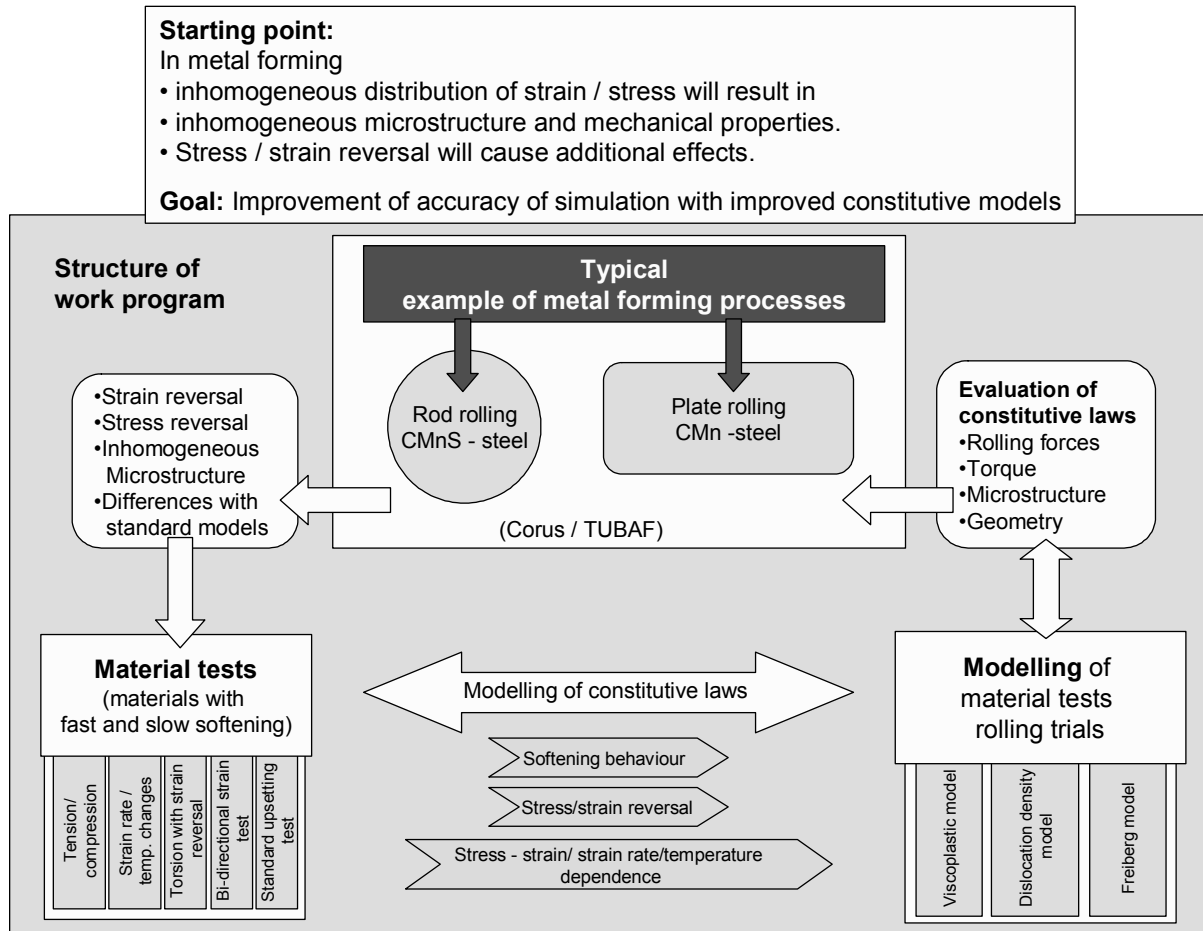


Fig. A5.1: Structure of work program

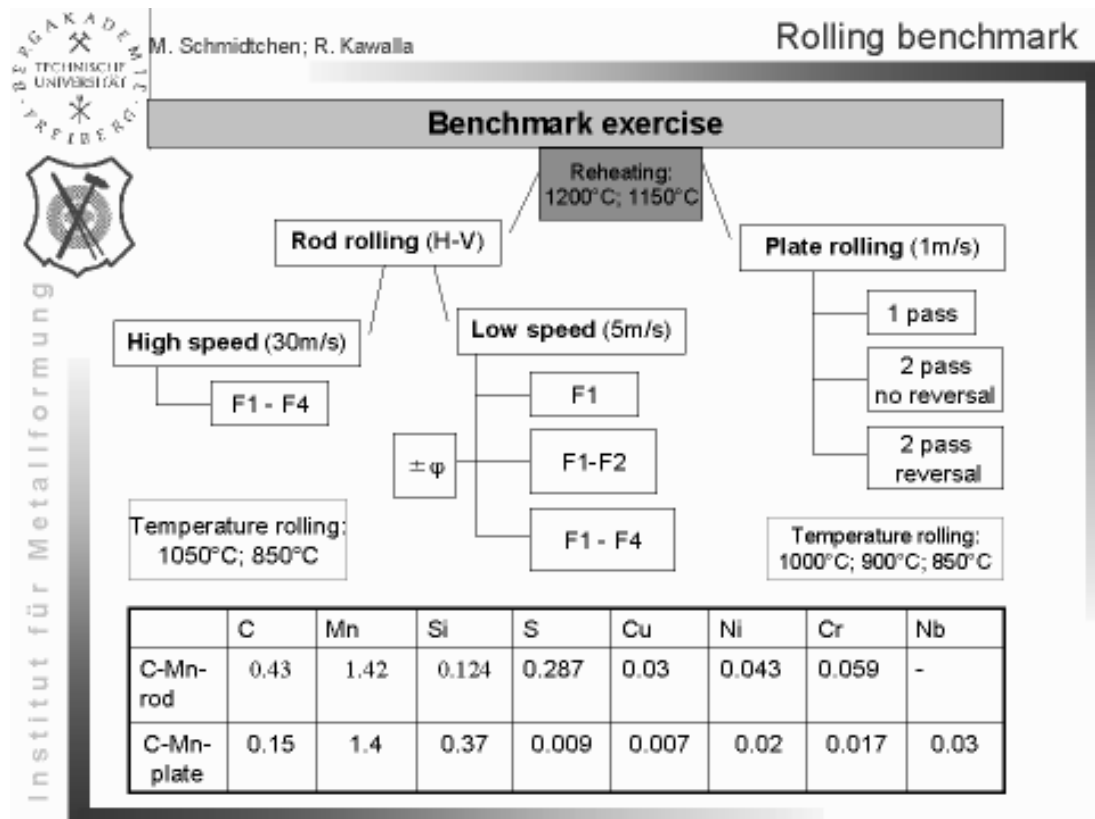


Fig. A5.2: Scheme of benchmark trials and steel grades used

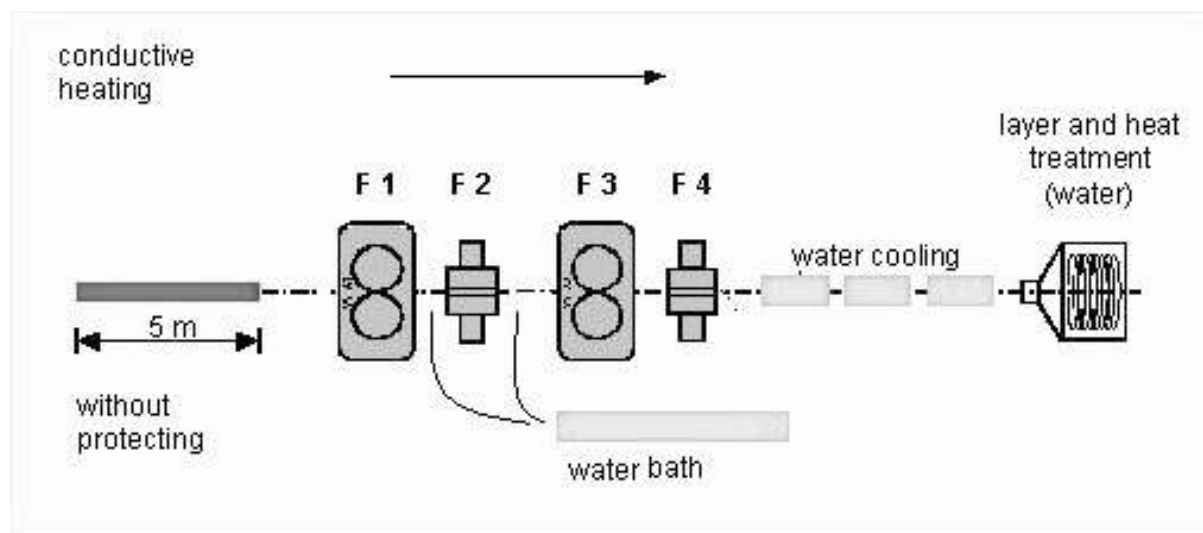


Fig. A5.3: Continuous high speed rolling mill
(water bath for trials F1-F2; F1 only; layer for high speed rolling only)

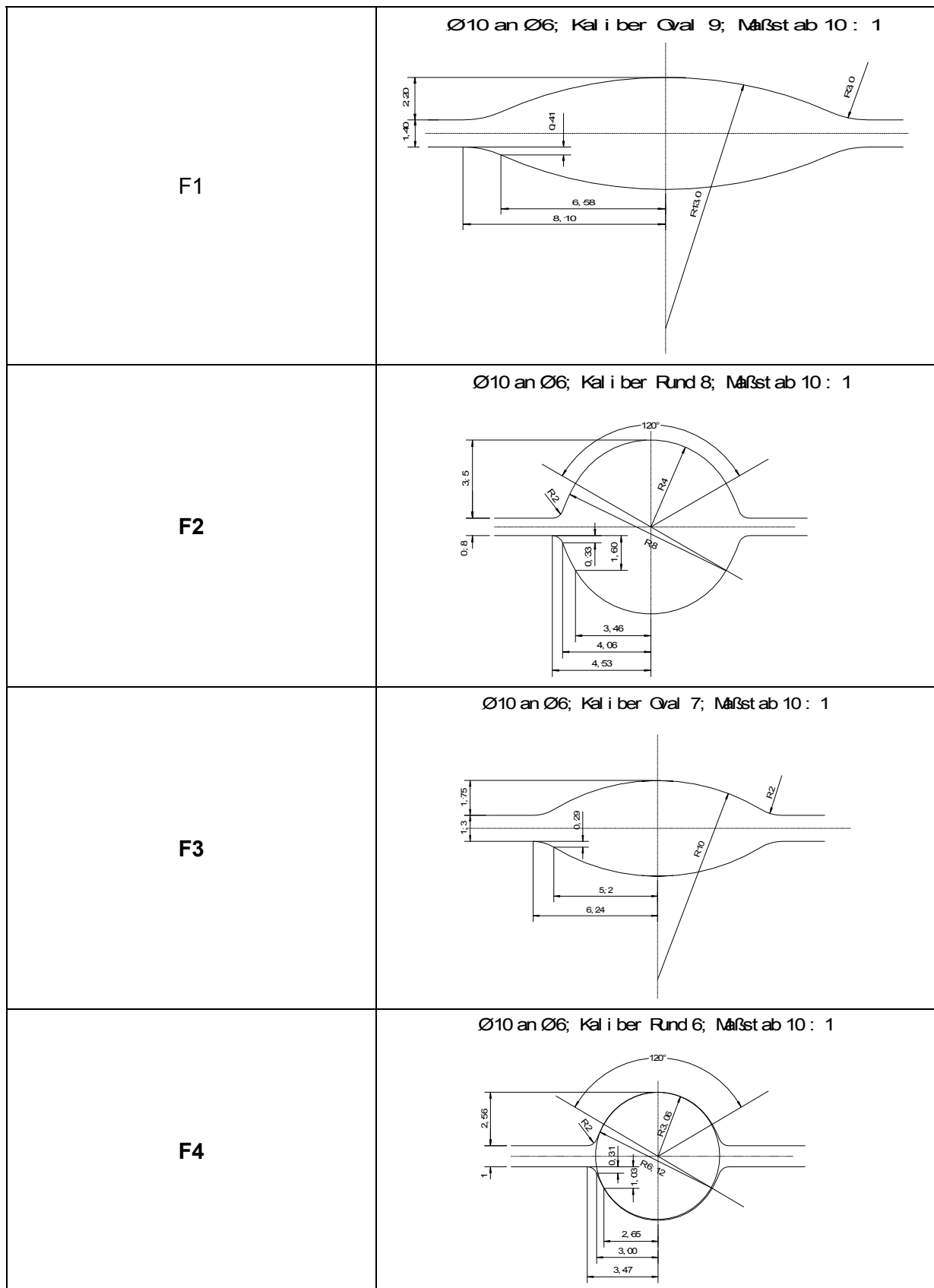


Fig. A5.4: Roll pass design

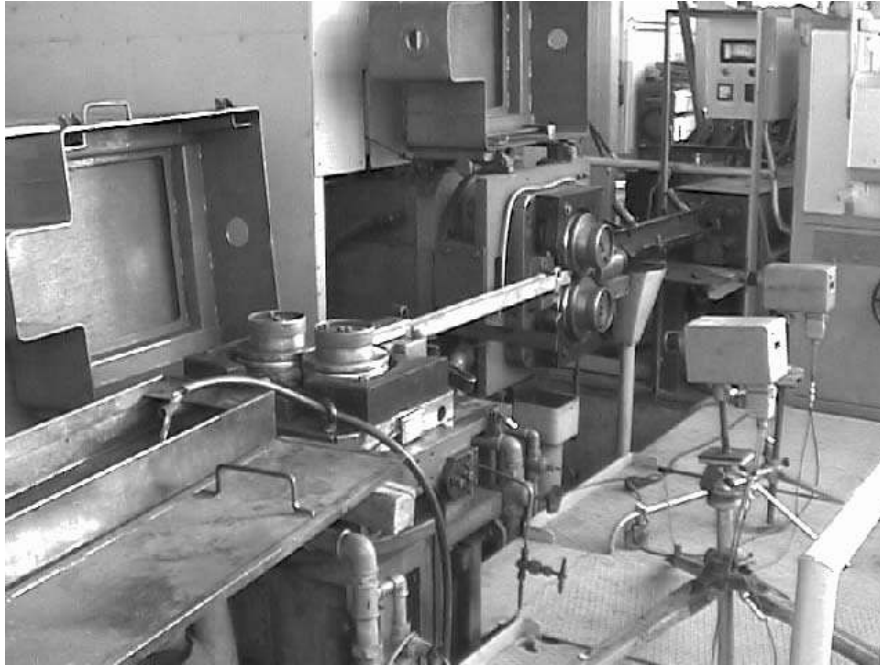


Fig. A5.5: Stands F1 (right) and F2 (left) for rolling F1-F2 at 5 m/s

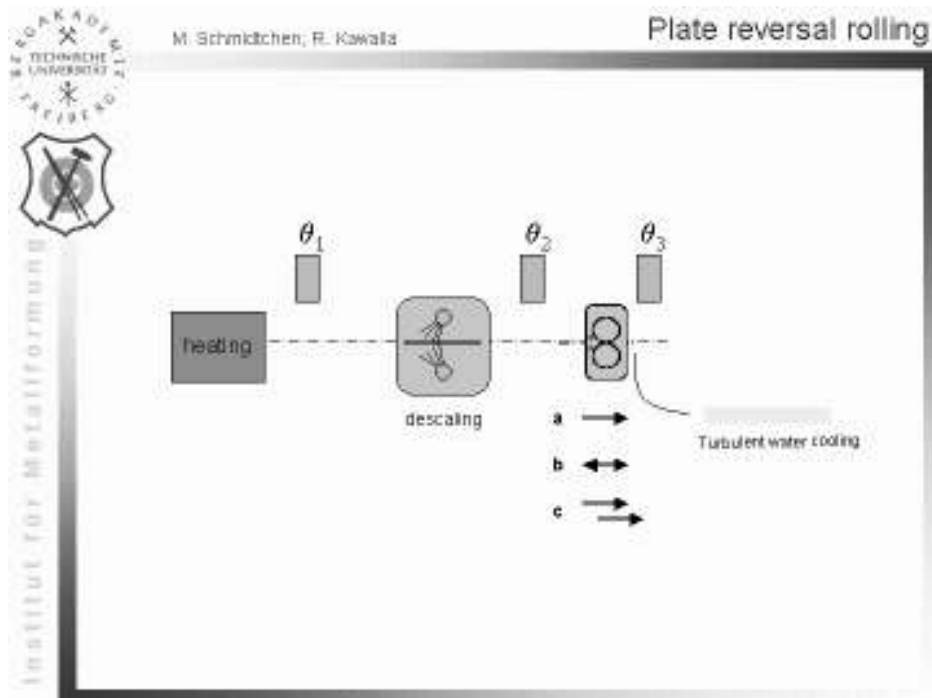


Fig. A5.6: Configuration of plate rolling trials
(a: single pass; b: reversal rolling; c: forward rolling 2 passes)

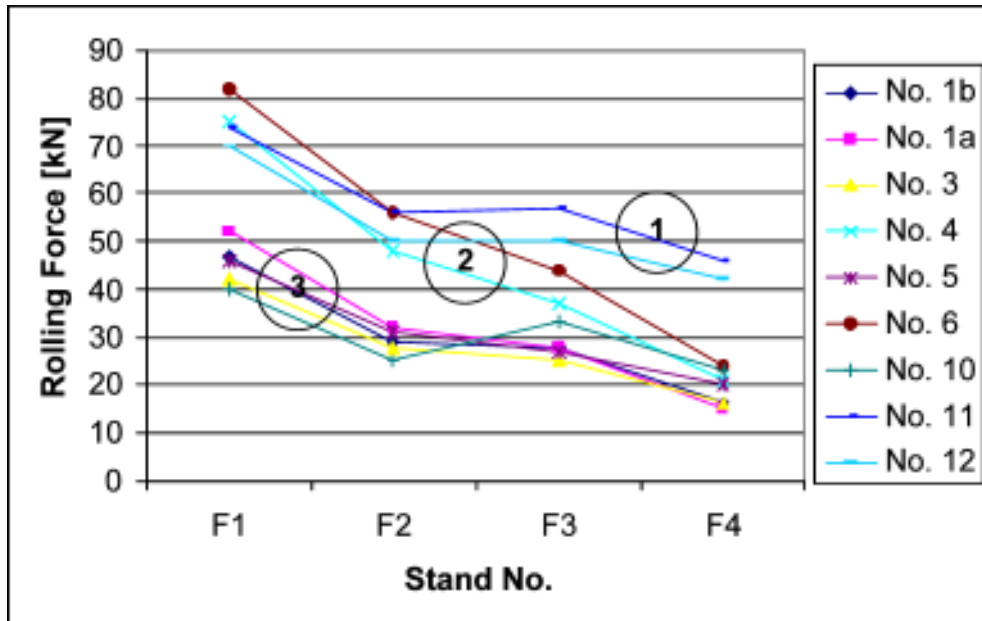


Fig. A5.7: Rolling forces during the continuous rod rolling (selection)

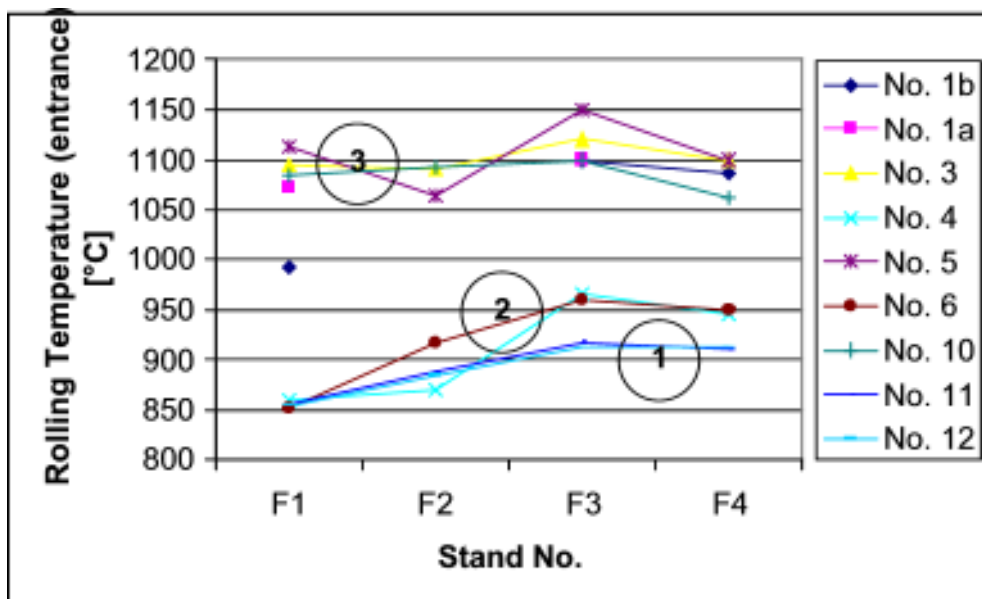


Fig. A5.8: Rolling temperatures during the continuous rod rolling (selection)

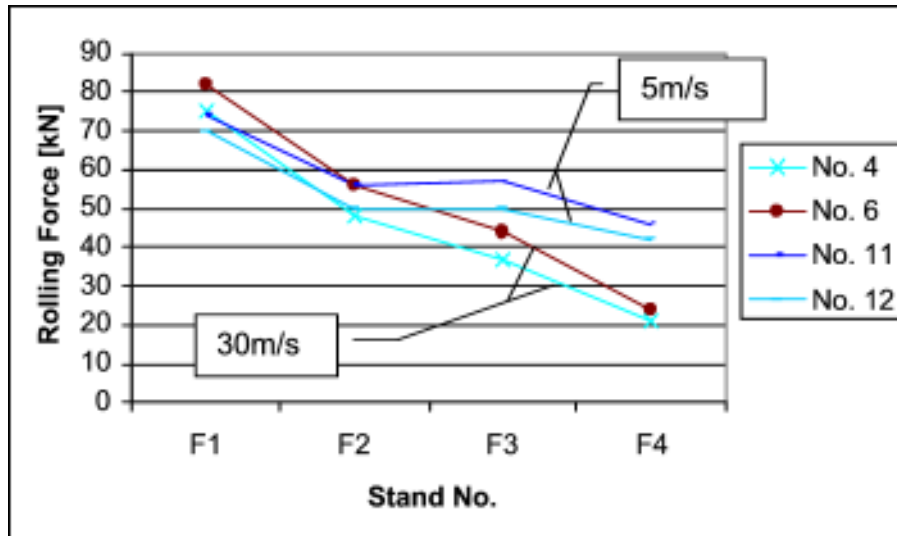


Fig. A5.9: Effect of reheating temperature and rolling speed on rolling forces during the continuous rod rolling – const. Temperature at entrance F1 – (Reheating temperature/time: 4 → 1230°C/15 s; 6 → 1150°C/60 s; 11 → 1150°C/60 s; 12 → 1190°C/60 s)

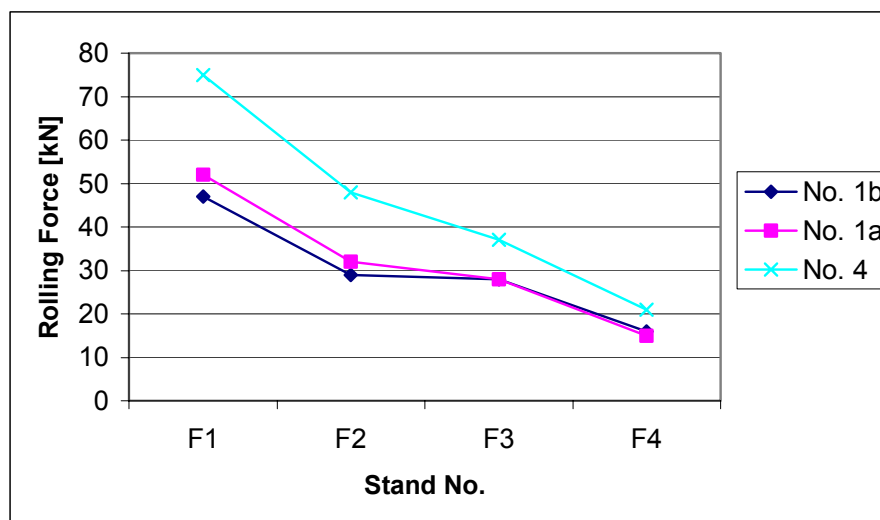


Fig. A5.10: Effect of temperature at the entrance of F1 on rolling forces during the continuous rod rolling at constant rolling speed (30 m/s)

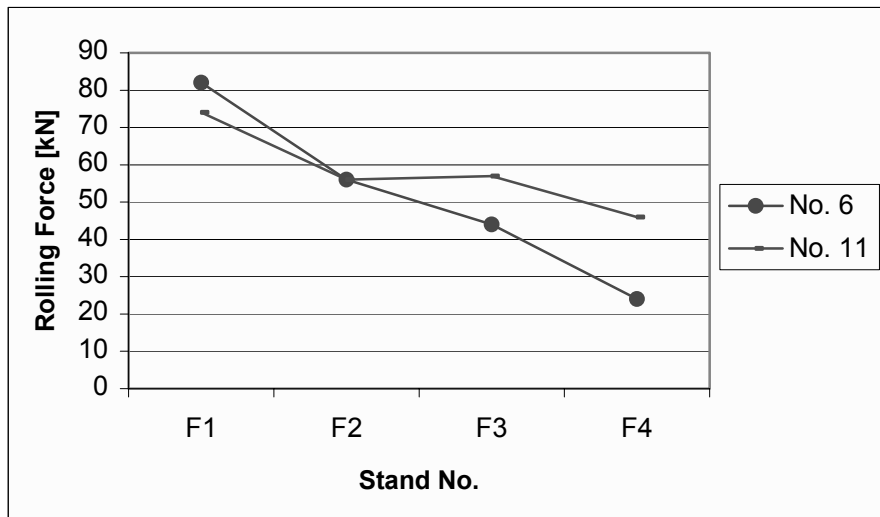


Fig. A5.11: Effect of rolling speed on rolling forces (const. Reheating temperature, reheating time and temperature at entrance F1) during the continuous rod rolling (6→30 m/2 s; 11 →5 m/s)

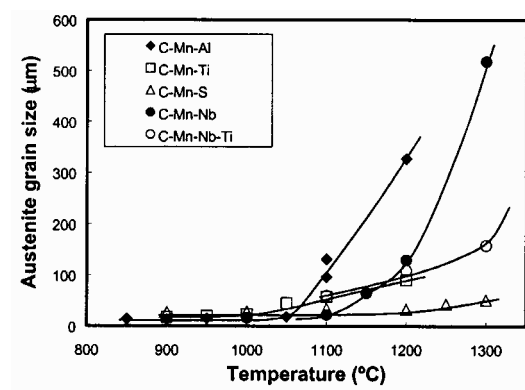


Fig. A5.13: Austenitic grain size evolution during heating for different steel grades [A5.4]

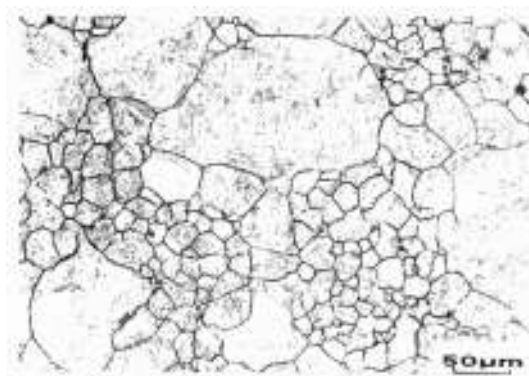
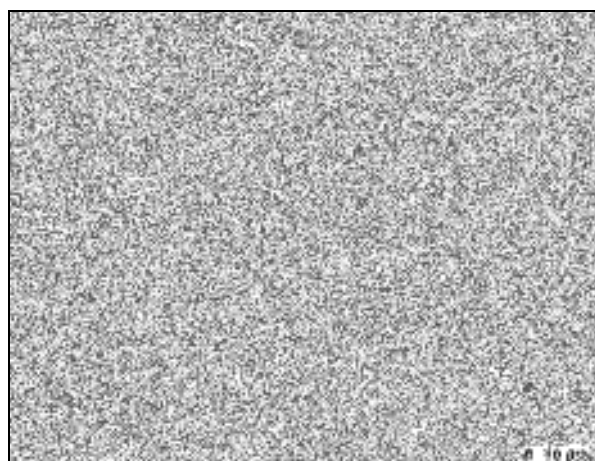
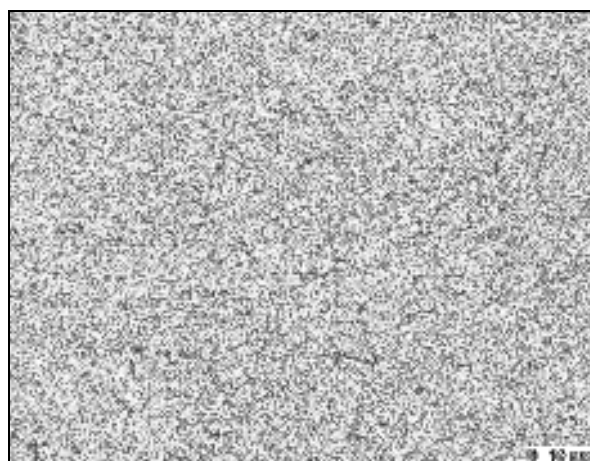


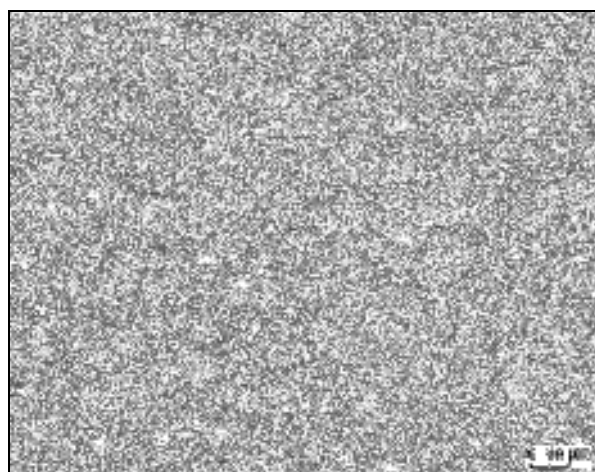
Fig. A5.14: Abnormal grain growth in a CMnNb grade; 20 min 1100°C [A5.4]



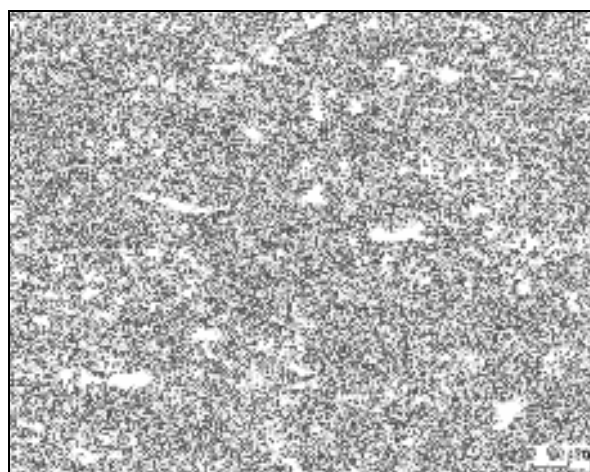
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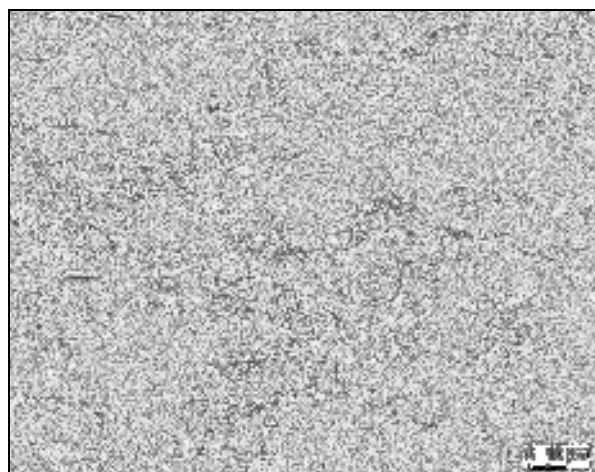
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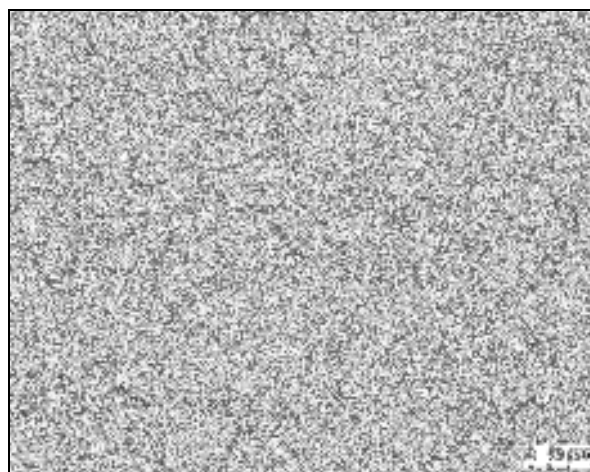
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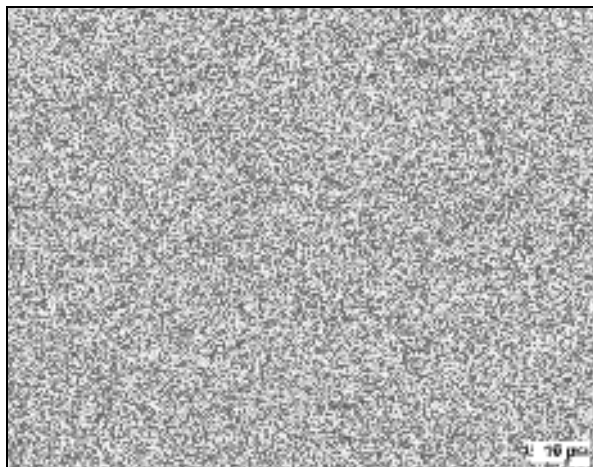
No. 4



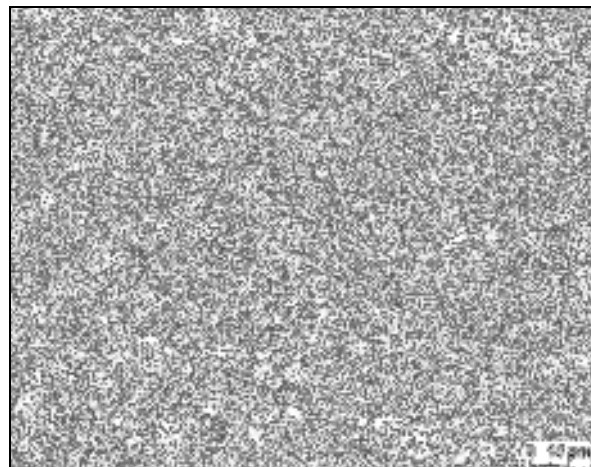
No. 5



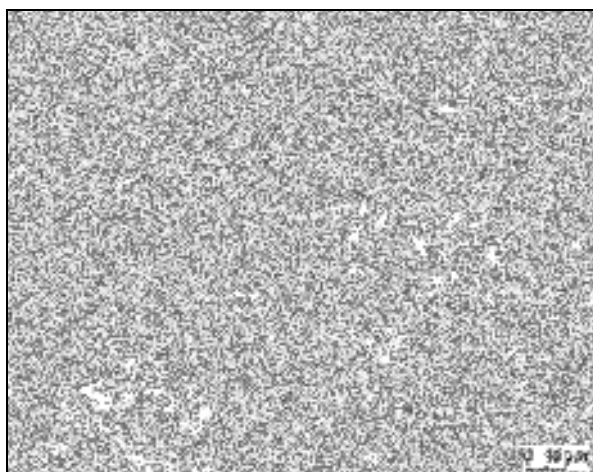
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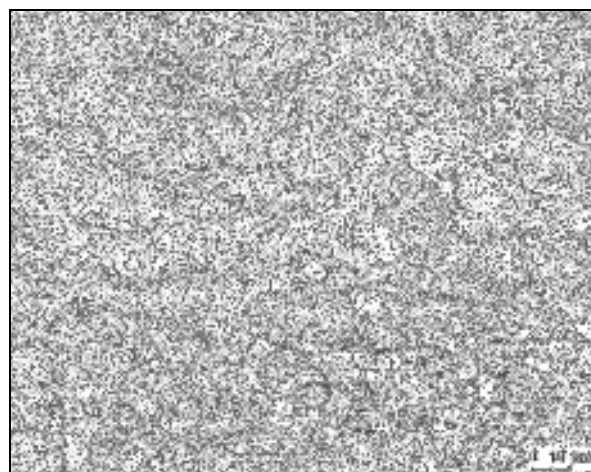
No. 11



No. 6



No. 10



No. 10

Fig. A5.12: Austenitic grain size at the end of pass F4 for different rolling conditions
(see [A5.1]; Fig. A5.3 and average grain size Table A5.2)

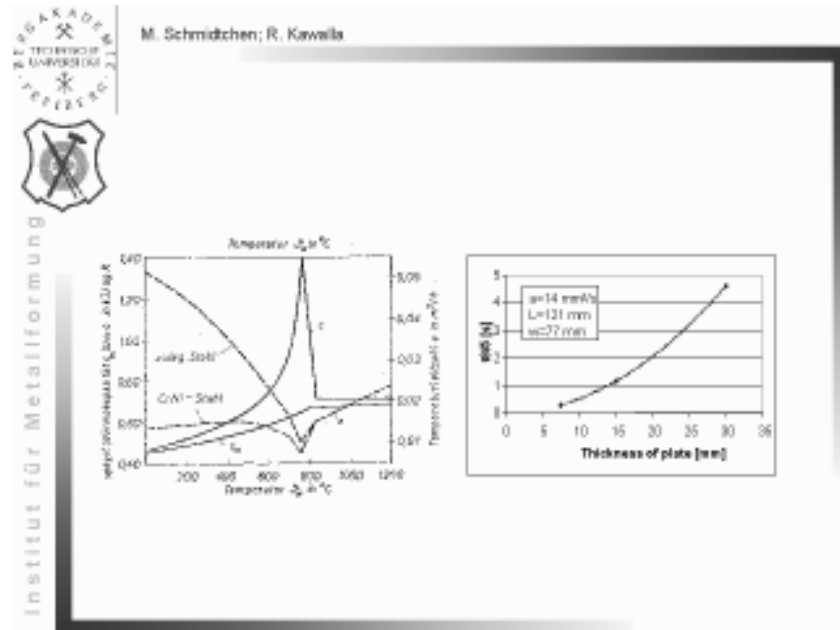


Fig. A5.15: Computed $t_{8.5}$ time for different thickness of plate

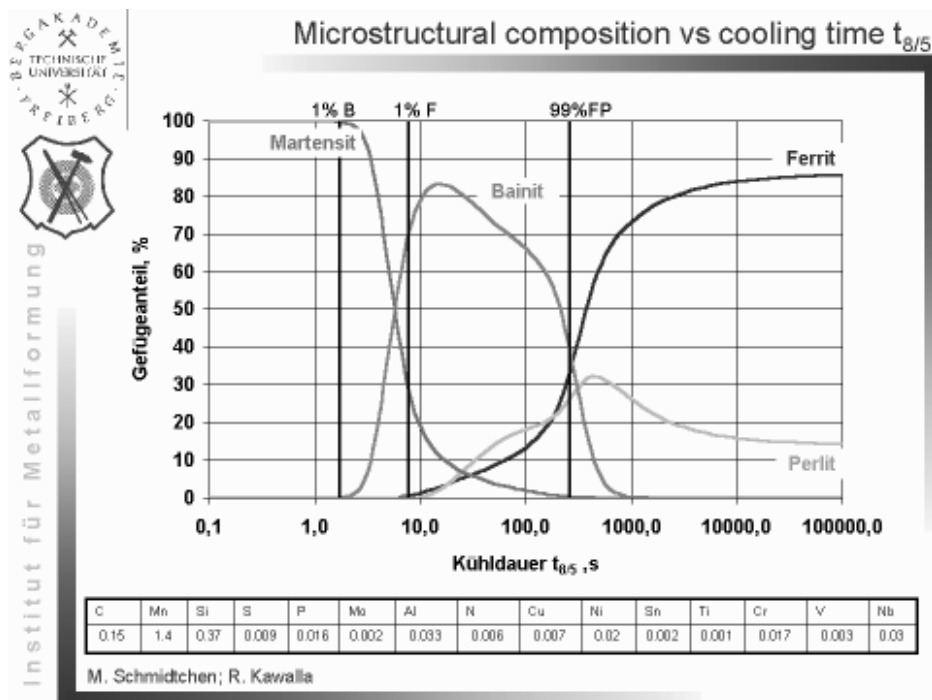


Fig. A5.16: Phase transformation behaviour of CMnNb steel

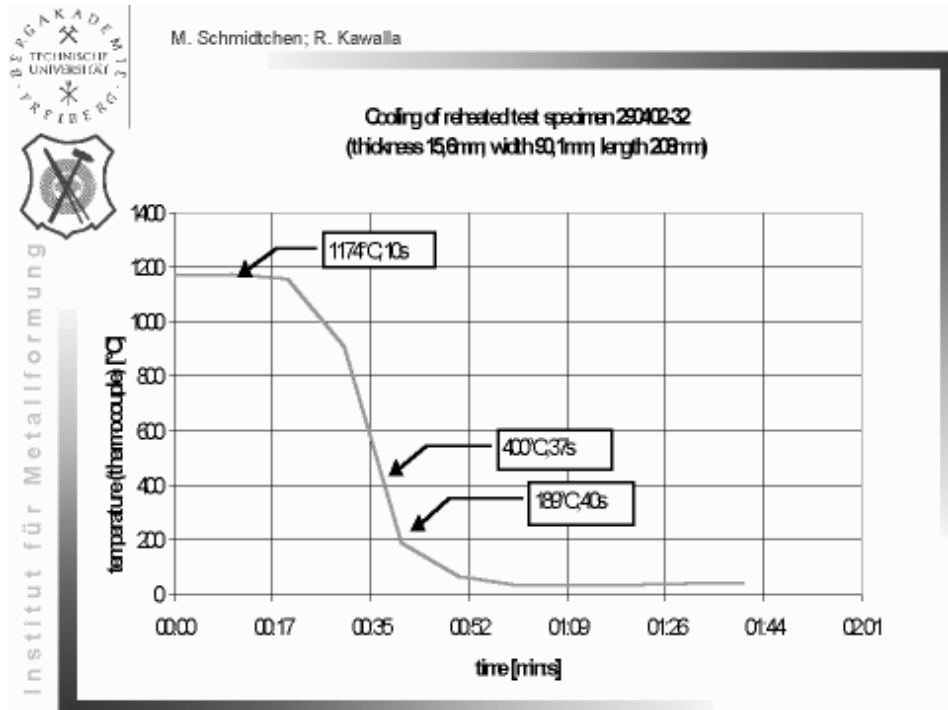


Fig. A5.17: Cooling test of plate rolling specimen thickness 15.6 mm

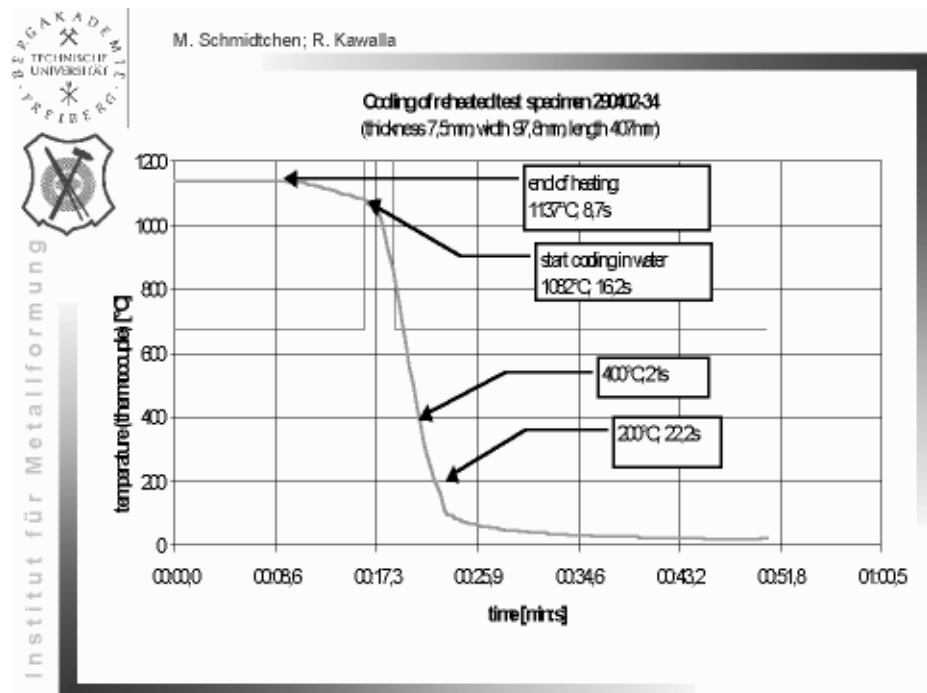


Fig. A5.18: Cooling test of plate rolling specimen thickness 7.5 mm

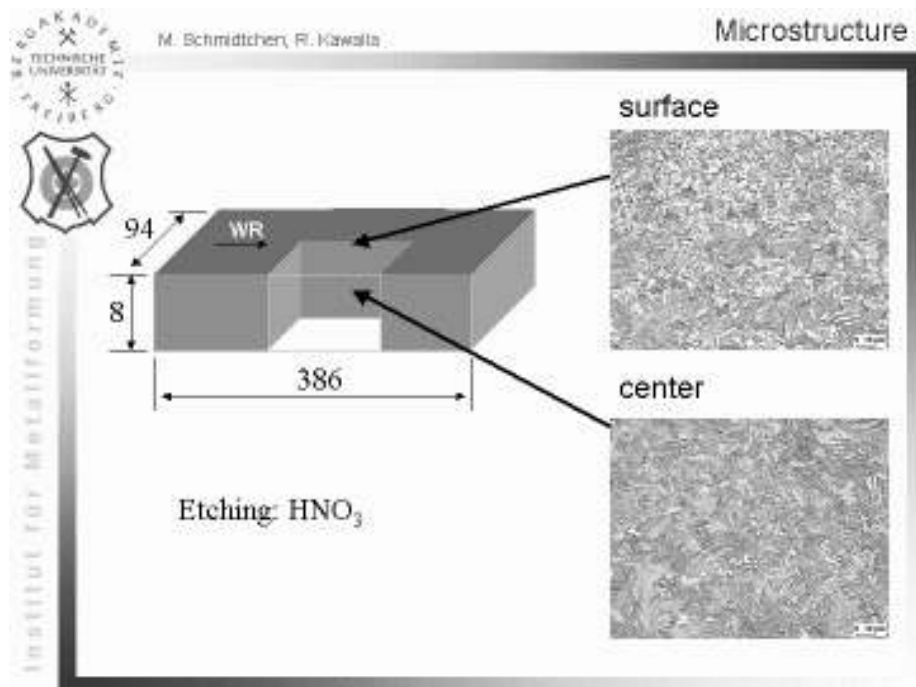


Fig. A5.19: Final geometry and microstructure of specimen in the 1 series (initial thickness 30 mm) after the 2 pass

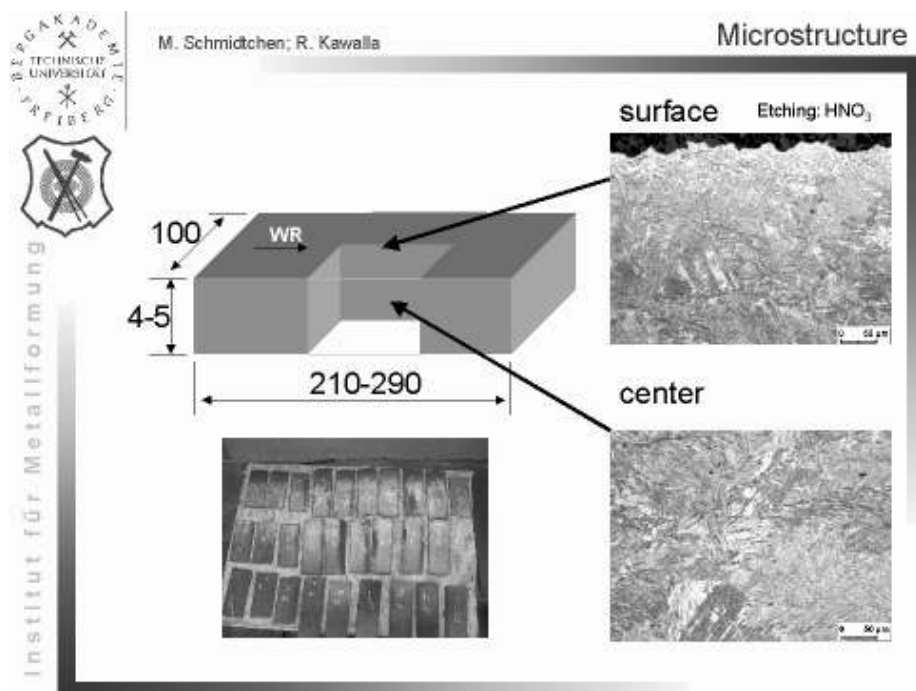
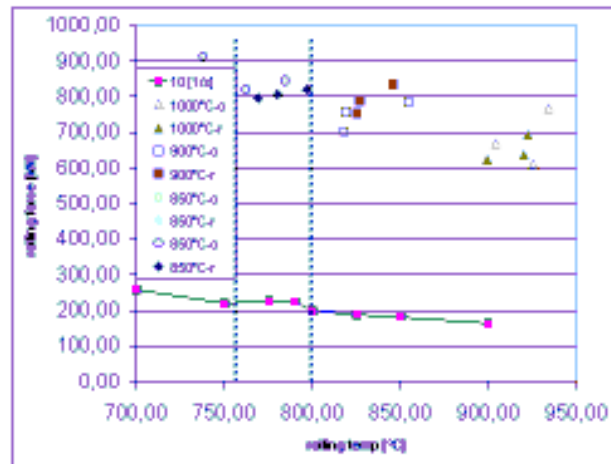
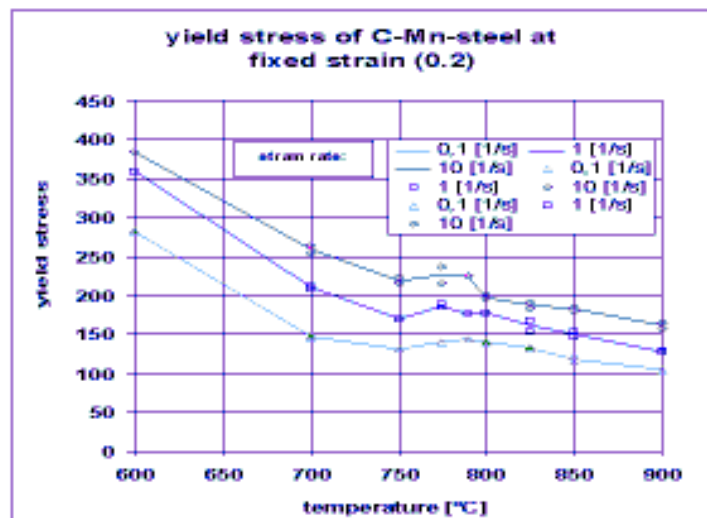


Fig. A5.20: Final geometry and microstructure of specimen in the 2 series (initial thickness 7 mm) after the 2 pass



Benchmark exercise

Fig. A5.21: Rolling forces and phase transformation behaviour (lower curve: yield stress 10/s; $\varphi = 0.2$; A_3 : 750°C ... 800°C)



Benchmark exercise

Fig. A5.22: Yield stress of CMnNb steel at fixed strain (0.2)

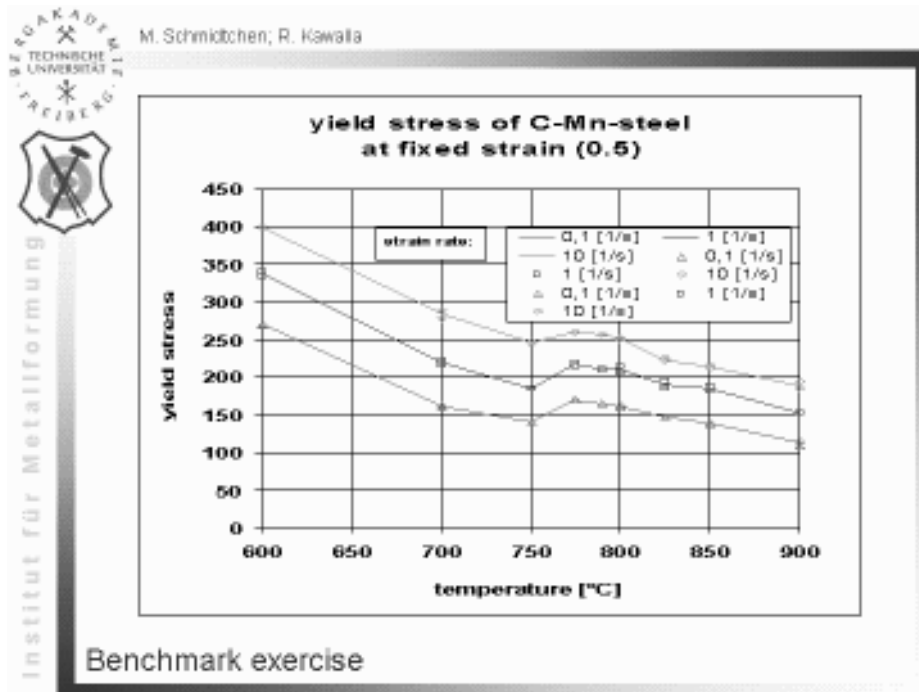


Fig. A5.23: Yield stress of CMnNb steel at fixed strain (0.5)

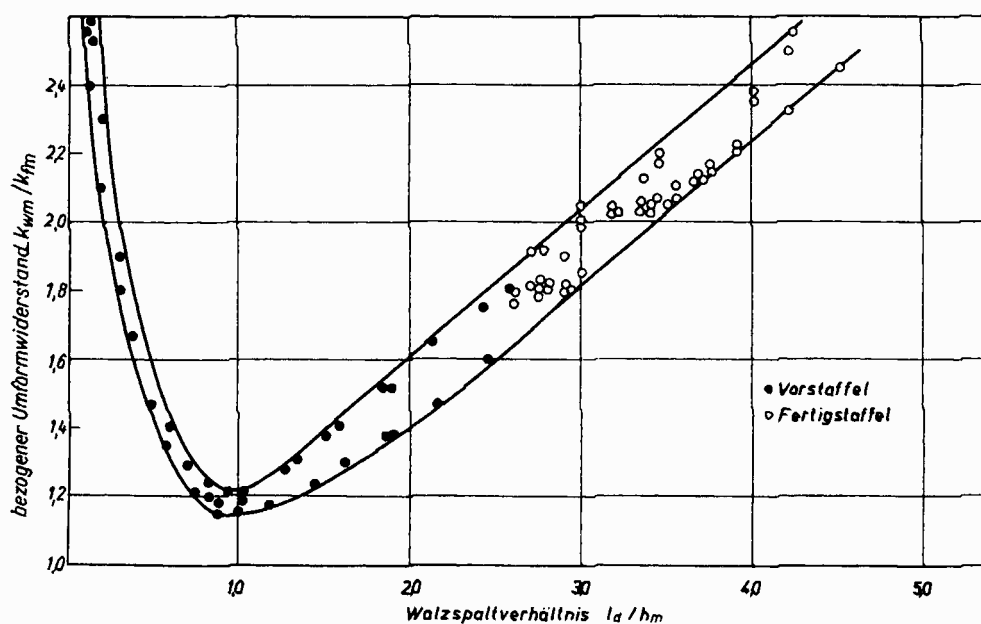


Fig. A5.24: Dimensionless forming resistance k_{Wm}/k_{Fm} v aspect ratio L_d/h_m [A5.5]

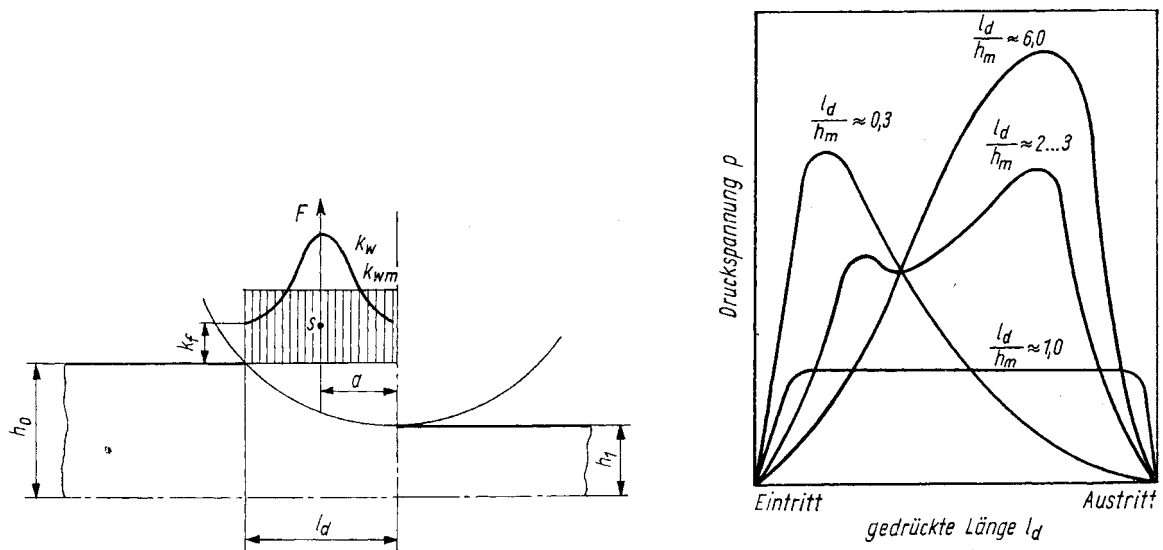


Fig. A5.25: Geometry for plate rolling (left) and its effect on normal pressure distribution (right) [A5.7]

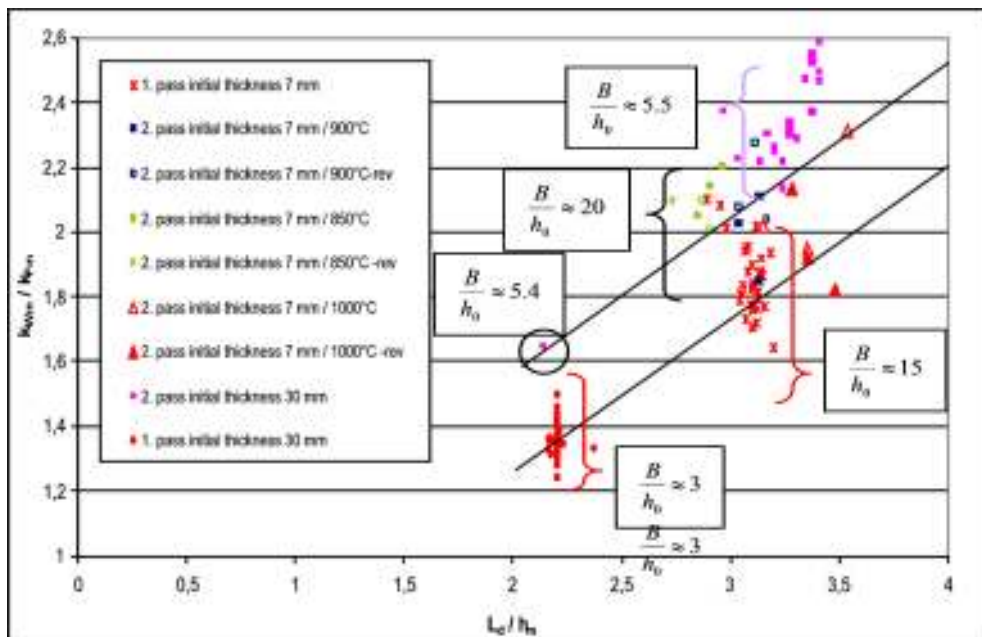


Fig. A5.26: Dimensionless forming resistance k_{wm}/k_{Fm} v aspect ratio L_d/h_m for 1 and 2 series of plate rolling

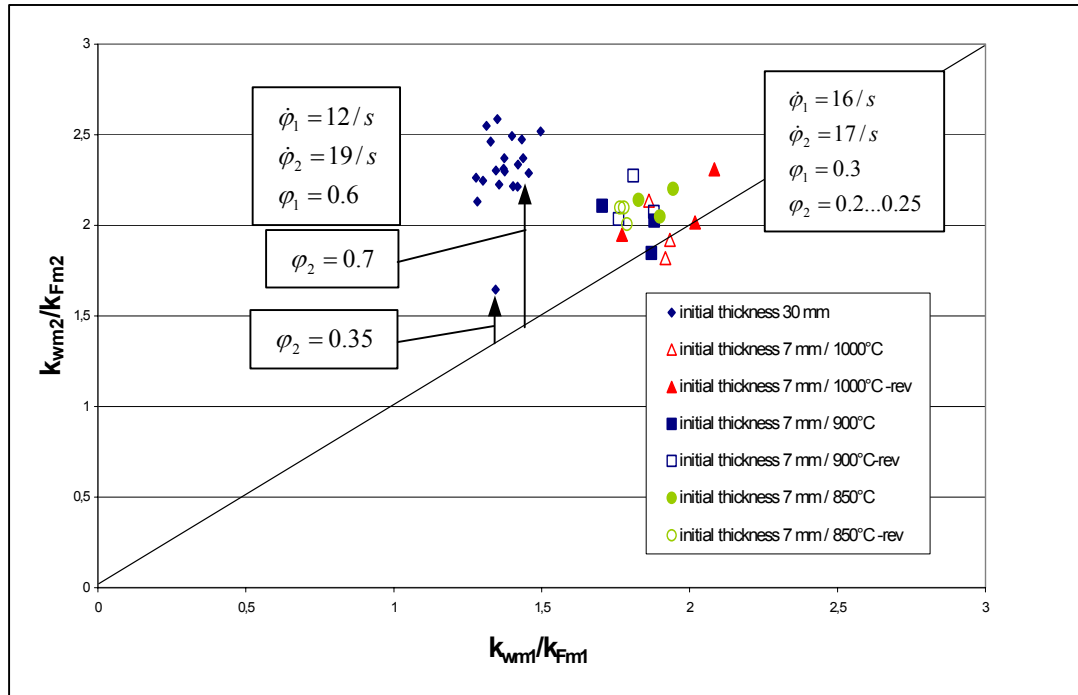


Fig. A5.27: Ratio of dimensionless forming resistance k_{wm1}/k_{Fm1} of the 1 pass v k_{wm1}/k_{Fm1} of the 2 pass (1 and 2 series)

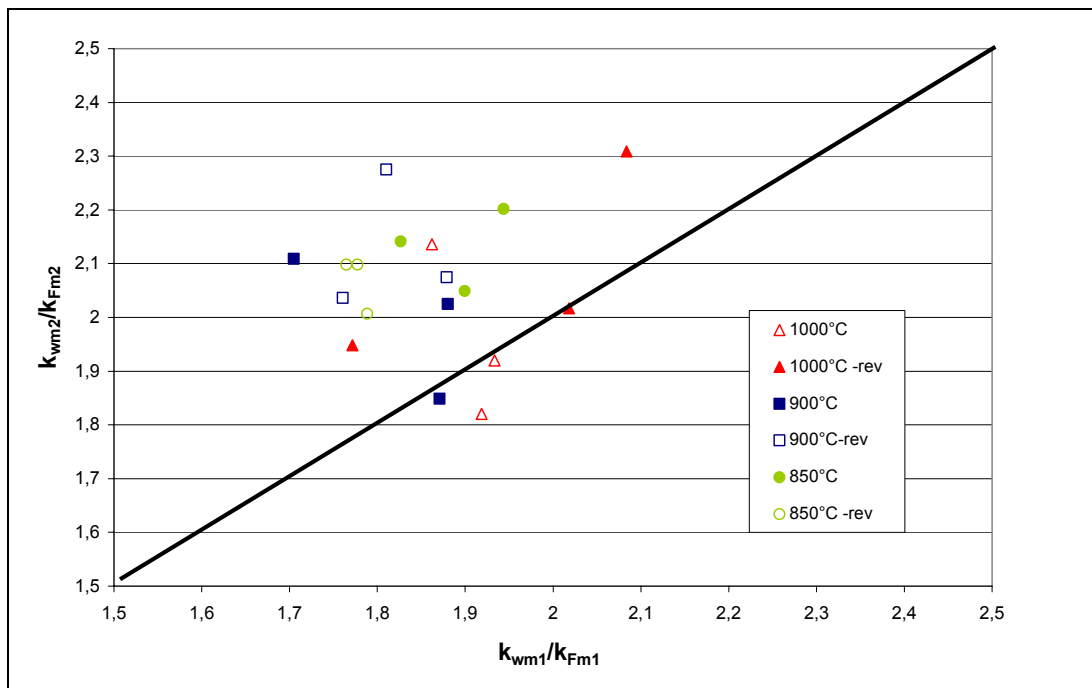


Fig. A5.28: Ratio of dimensionless forming resistance k_{wm1}/k_{Fm1} of the 1 pass v k_{wm1}/k_{Fm1} of the 2 pass (2 series (7 mm) only)

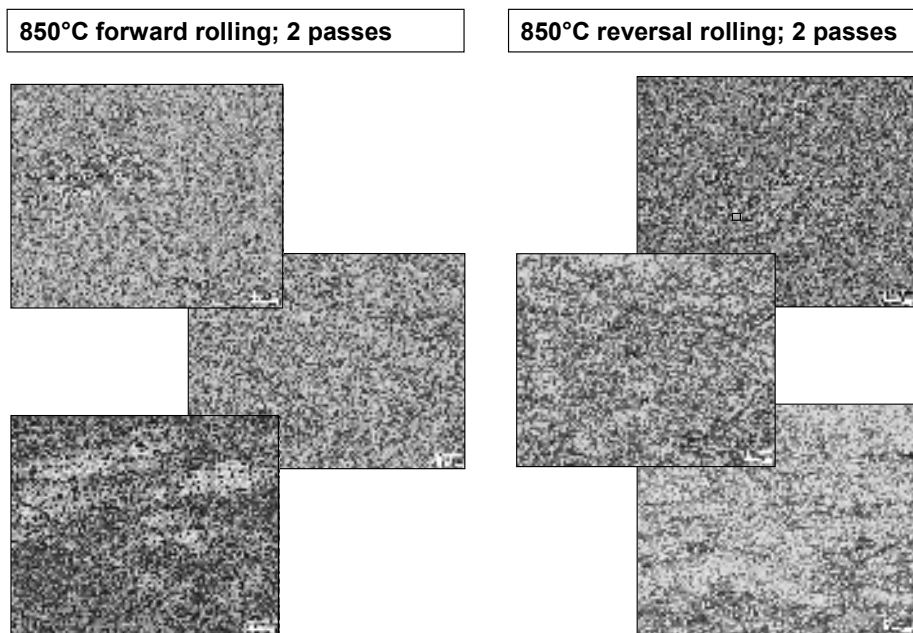


Fig. A5.29: Microstructural evolution during 2 passes in dependence on the rolling direction

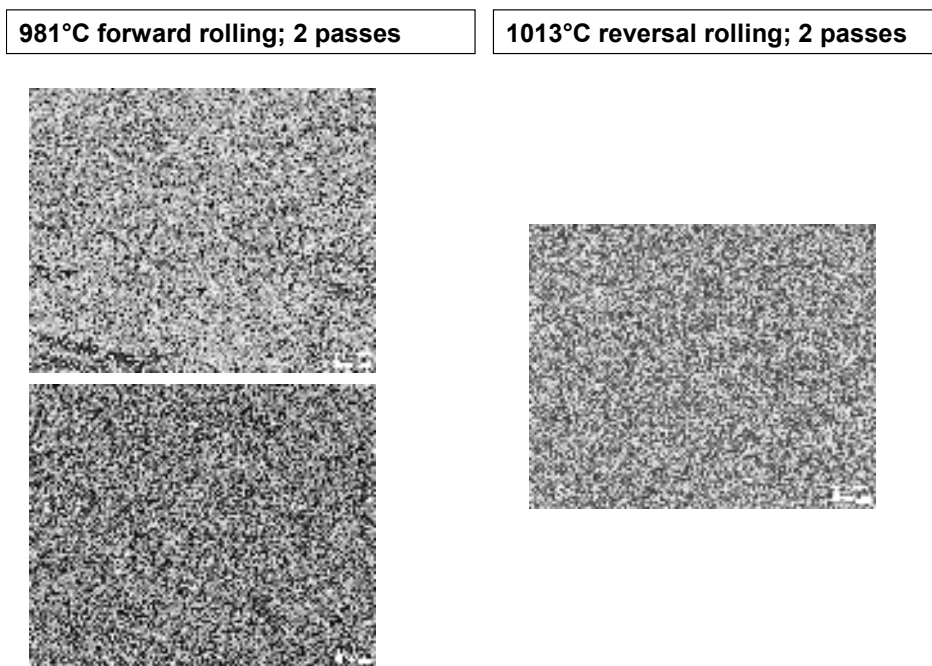


Fig. A5.30: Microstructural evolution during 2 passes in dependence on the rolling direction

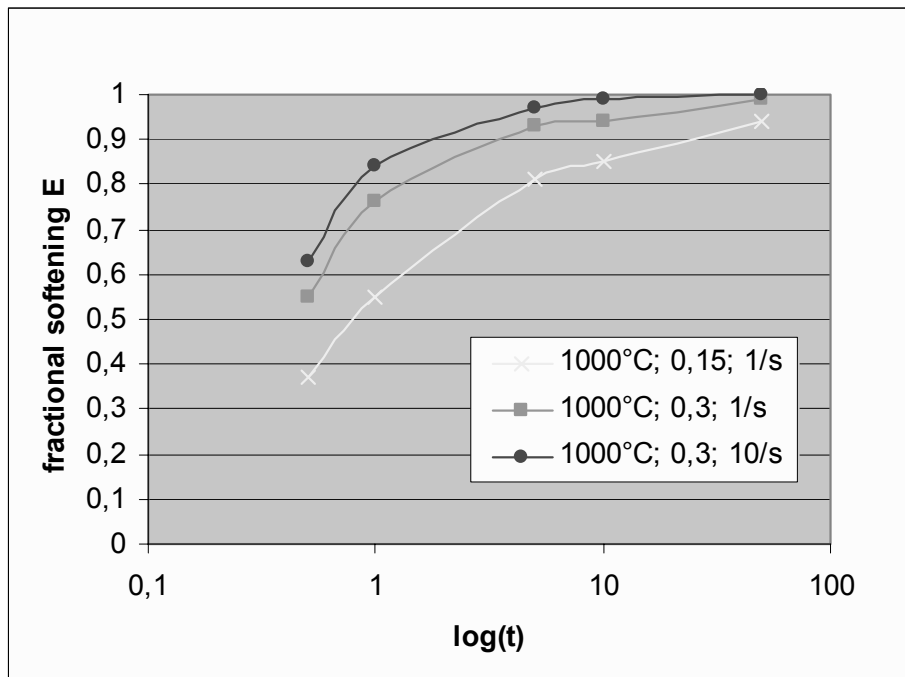
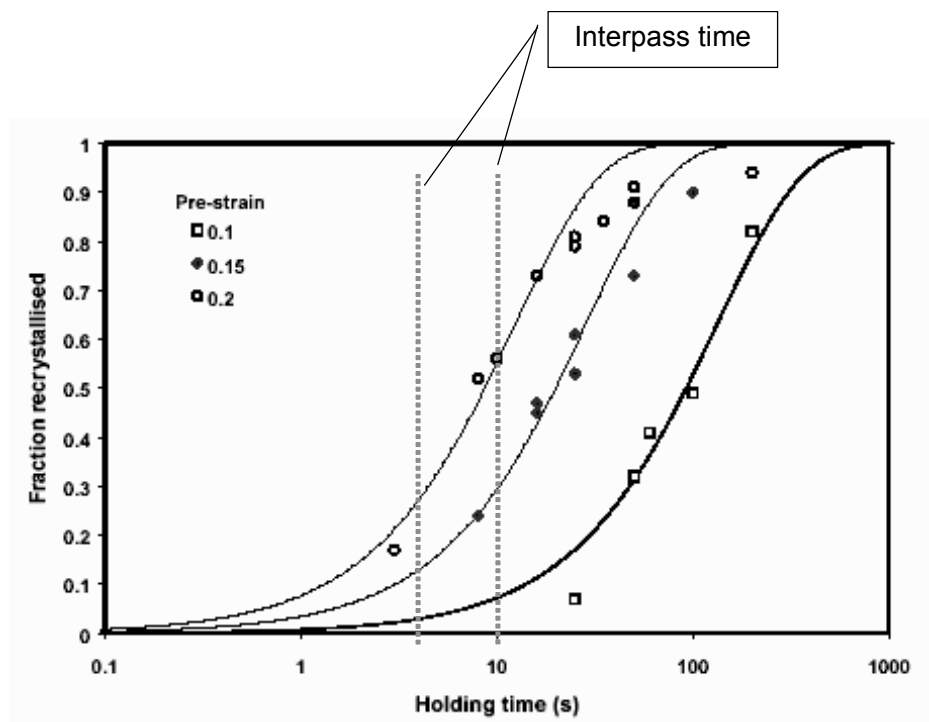


Fig. A5.31: Fractional softening for monotonic compression test: 1000°C



From CEIT

Fig. A5.32: Fractional softening for monotonic test: 0.1/s; 1150°C after CEIT

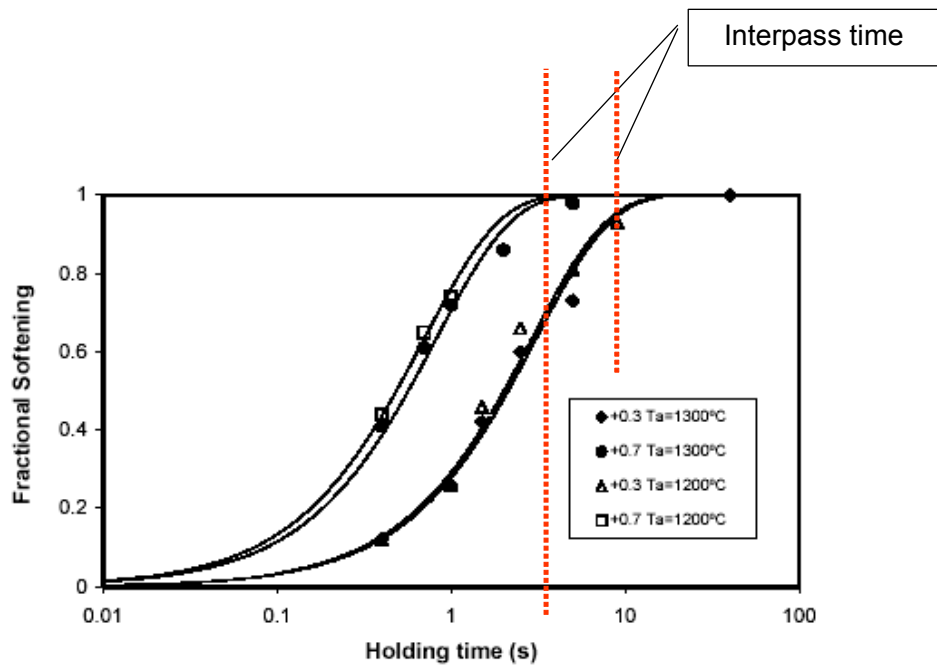


Fig. A5.33: Fractional softening of monotonic test: 1/s; 1100°C after CEIT

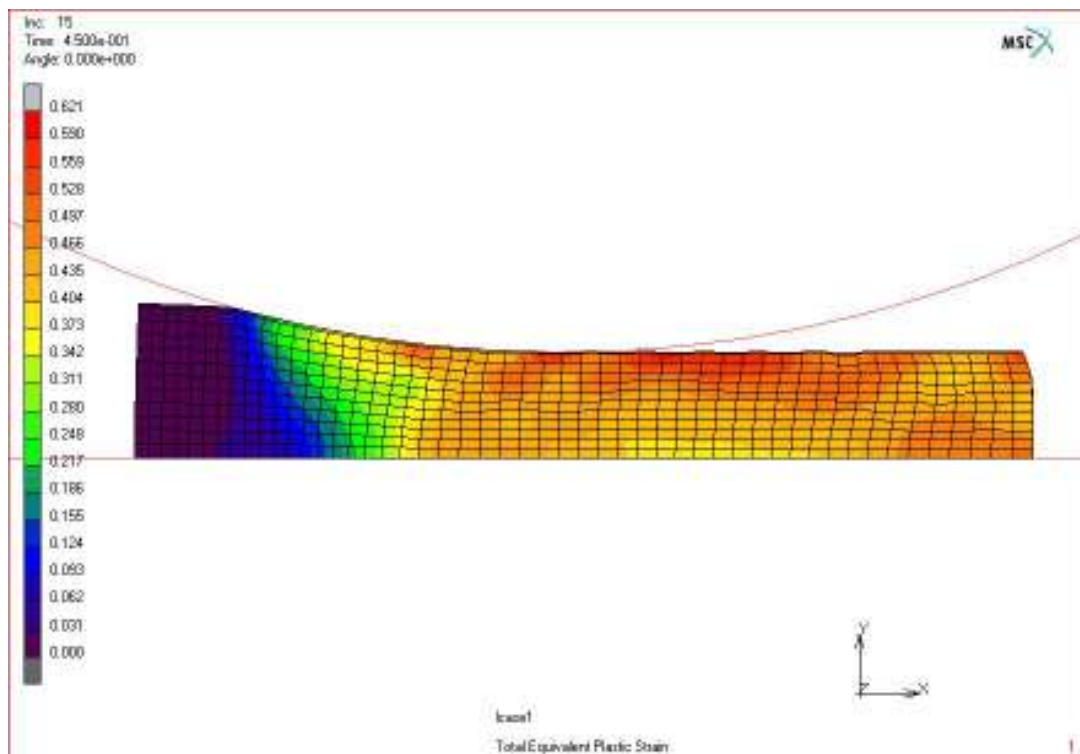


Fig. A5.36: FE computations: total equiv. plastic strain and flow lines for the rolling trial W3 (forward rolling: steel grade C15, 45 mm $\rightarrow$ 31 mm, 0.25 m/s 1100°C) – centre line is symmetry line

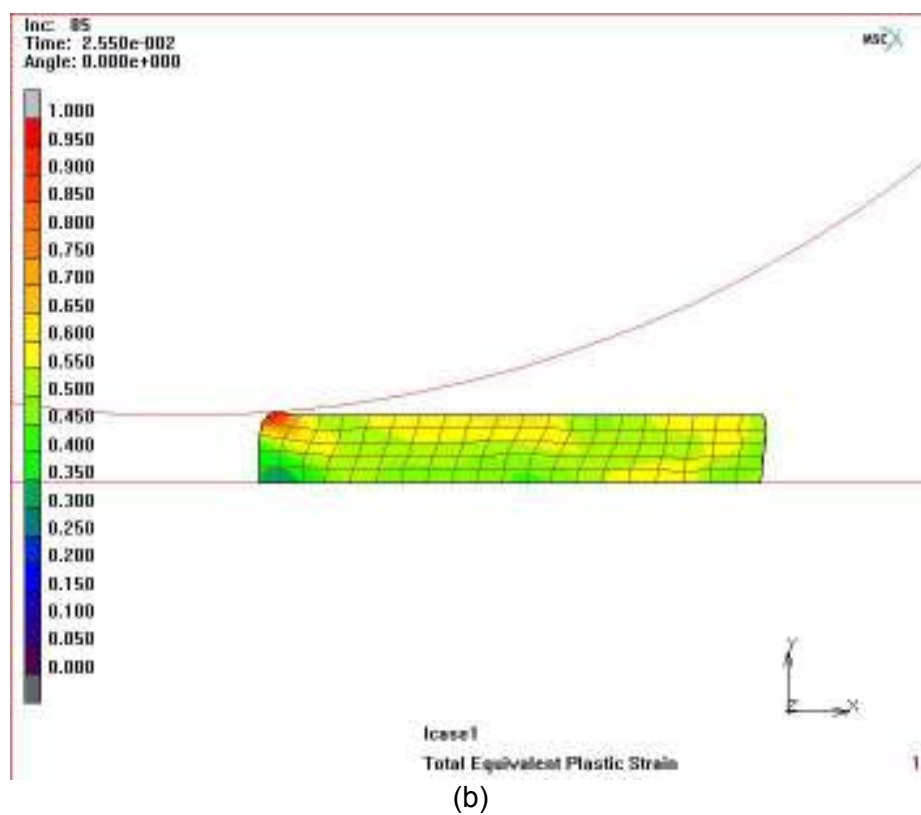
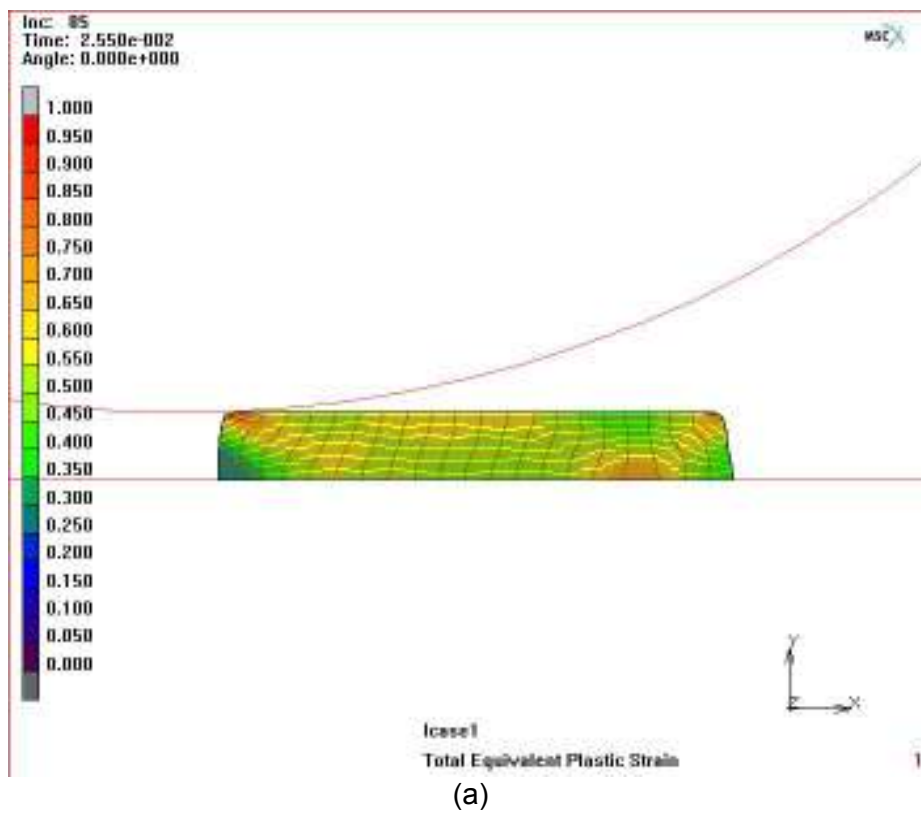


Fig. A5.34: FE computations: total equiv. plastic strain and flow lines for the 1st pass of forward rolling (a) and reverse rolling (b)

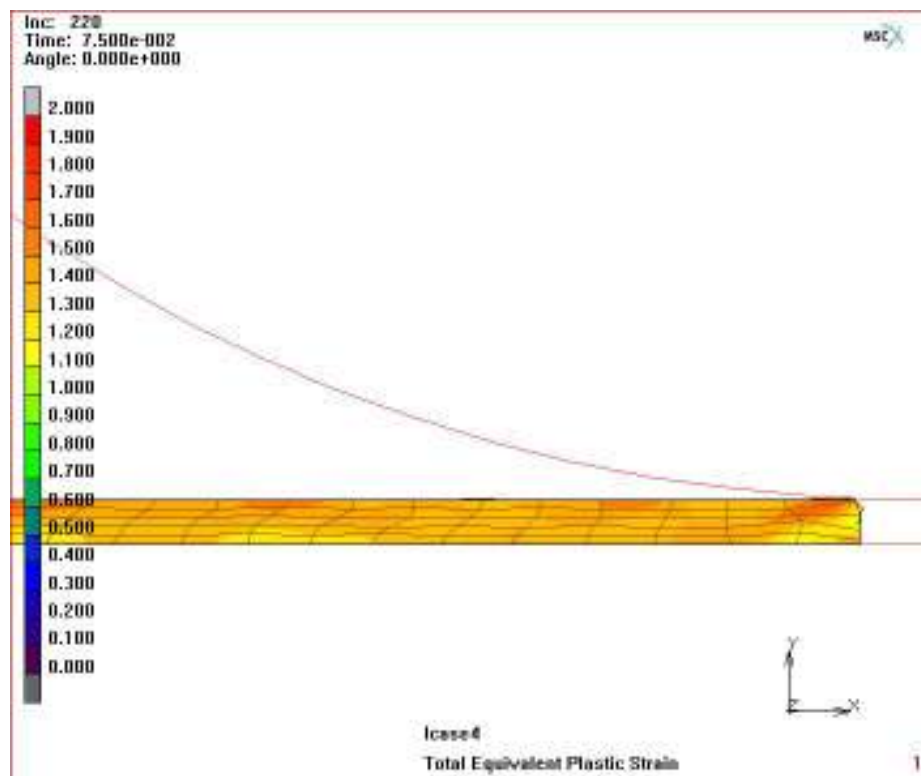
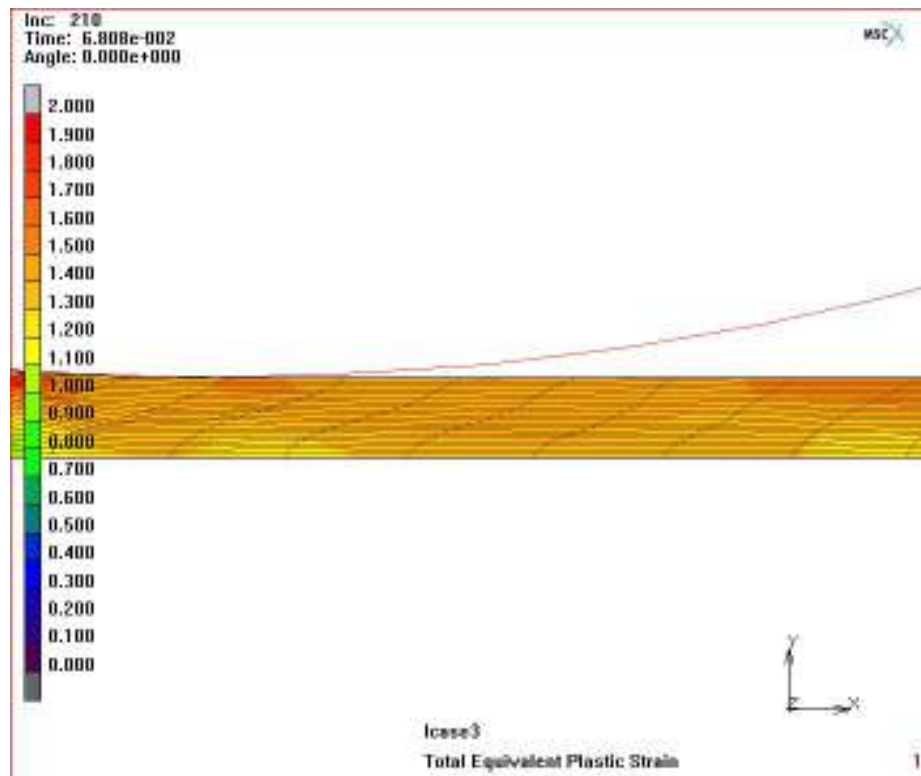


Fig. A5.35: FE computations: total equiv. plastic strain and flow lines for the 2nd pass of forward rolling (a) and reverse rolling (b)

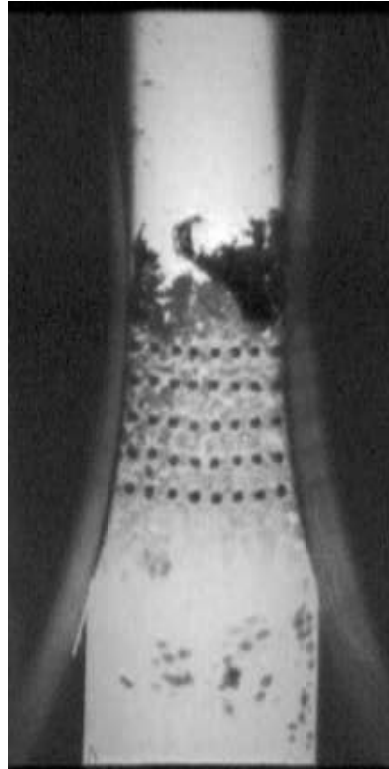
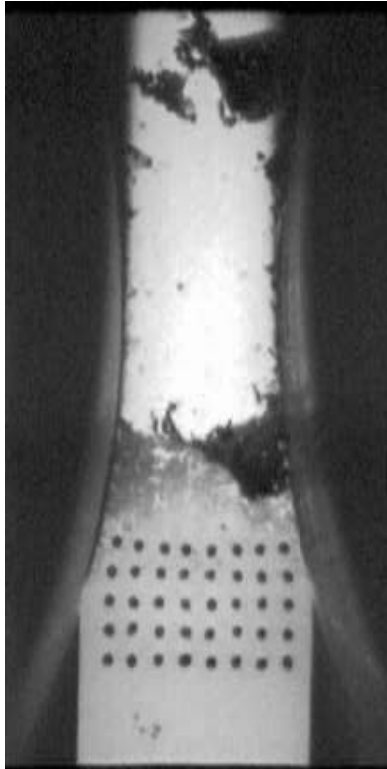
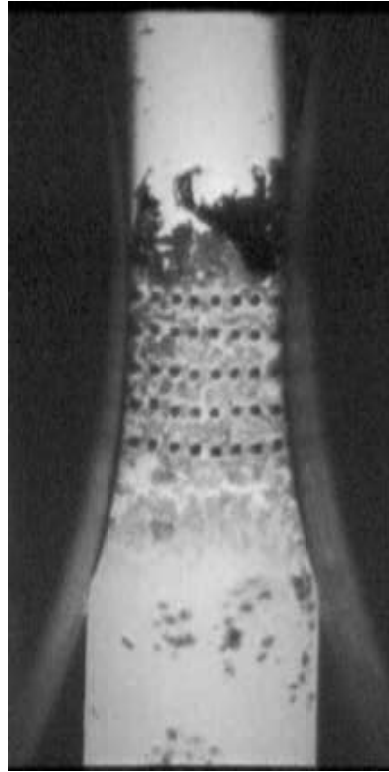
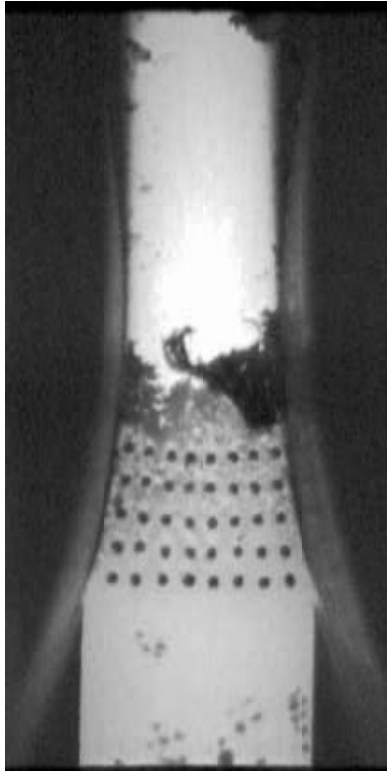


Fig. A5.37: Rolling trial W3 with marked surface for viscoplasticity investigations
(forward rolling: steel grade C15, 45 mm $\rightarrow$ 31 mm, 0.25 ms, 1100°C)

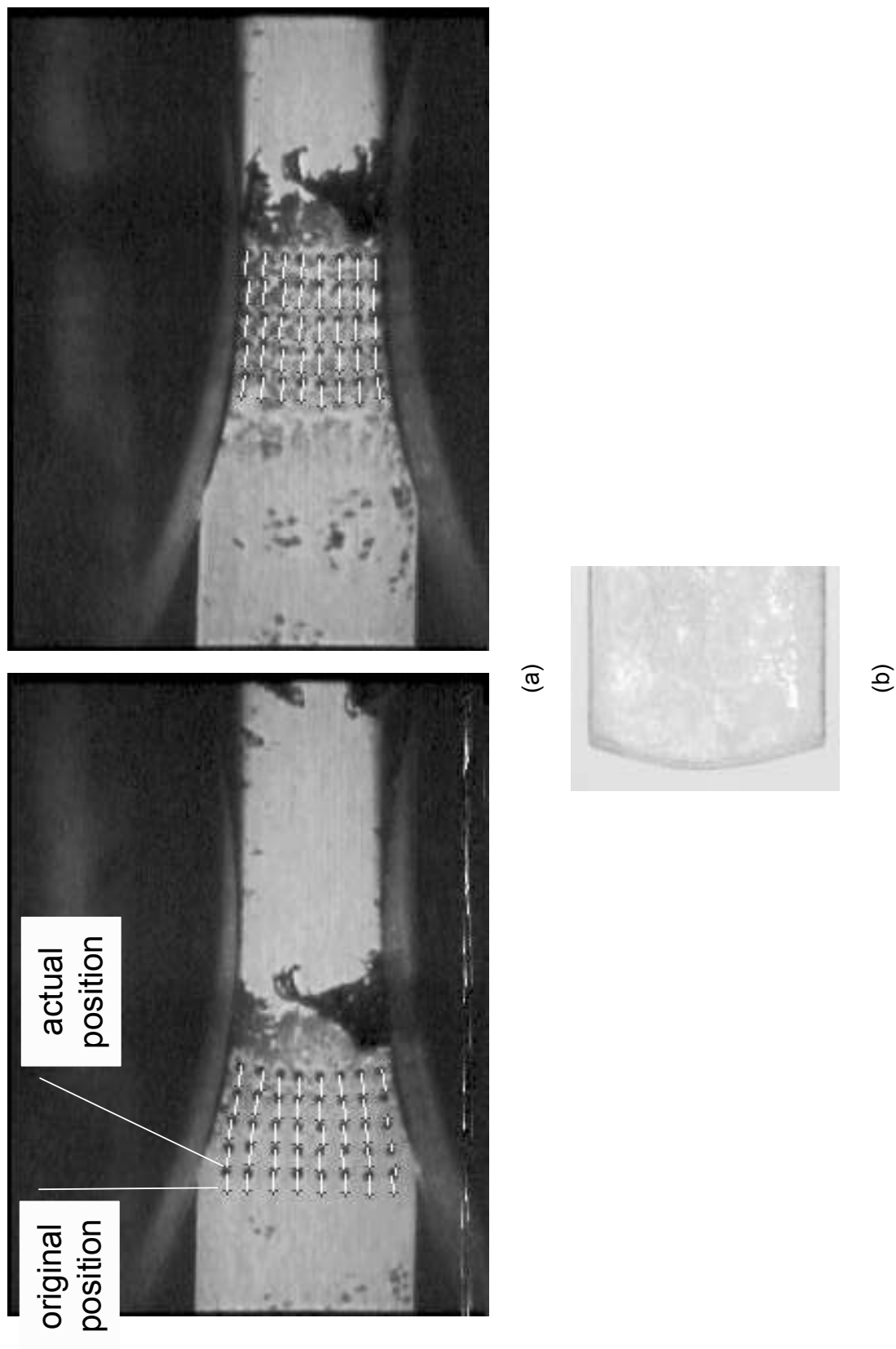


Fig. A5.38: Automatically obtained displacement increments for two positions during the rolling trial W3 with marked surface (a) and cross section of the rolled specimen (b) (forward rolling: steel grade C15, 45 mm $\rightarrow$ 31 mm, 0.25 m/s, 1100°C)

Appendix 6

Detailed Report by the University of Oulu Authors: L P Karjalainen and M C Somani

SUMMARY

In order to assess the importance of precise deformation conditions used in modelling of hot working processes, the effects of temperature and strain rate histories as well as that of the reversed deformation on the flow stress and recrystallisation behaviour of both the austenite and ferrite phases have been investigated in various steels.

According to the experimental results, the temperature history seems to have only an insignificant influence on the flow stress and recrystallisation rate of austenite. In microalloyed steels, during a temperature drop to a low temperature, precipitation can occur that may prevent the subsequent recrystallisation. However, in the rolling gap the surface temperature drops are very short in duration so that precipitation can hardly occur to any significant extent.

Under abruptly decreasing strain rate, in the dynamic recovery regime, the final flow stress remains higher than that at the corresponding constant final strain rate, close to the stress at the constant mean strain rate of the whole deformation history. This means that the mean flow stress is higher under the varying strain rate than at constant mean strain rate. The static recrystallisation rate lags even more than the flow stress so that it has a certain memory from the previous strain rate history. Following an abrupt decrease in the strain rate, the recrystallisation rate remains close to that at the high strain rate before the drop. However, because the effect of strain rate on the recrystallisation kinetics is weak, the effects are not very significant and the mean strain rate can be used for modelling with reasonable accuracy. The composition of the steel has no influence on that.

Reversing the direction of deformation from tension to compression results in the Bauschinger effect in the flow stress of austenite, with the magnitude order of 5%. The reversal leads to a retarded recrystallisation rate, the amount depending on the strain after the reversal (at a constant strain before the reversal). The rate is decreased as much as by 2-3 times, as the second strain is 0.02-0.04 and the first strain 0.2. Qualitatively this is consistent with observations from torsion tests. Decreasing the strain rate in the final stage of deformation after the reversal causes an additional retardation. Partial recrystallisation after the first deformation before reversing the strain enhances the recrystallisation, but the effect is not pronounced. No distinct effects of the reversal on the recrystallised austenite grain size were detectable.

In the ferrite phase, the influence of strain rate change on the flow stress level and the recrystallisation kinetics was found to be qualitatively similar to that in the austenite phase. Strain reversal was found, by the double-compression tests and by optical microscopy, to retard the recrystallisation of ferrite, and qualitatively its effect was similar as for the austenite. However, a pronounced coarsening of the ferrite grain size was observed at small reversed strains.

After changing the strain rate abruptly, the cell size formed in the ferrite seemed to be larger than at constant high strain rate revealing that the driving force for the static recrystallisation decreases rapidly due to the drop of the strain rate. The misorientation distribution of the

grain boundaries was not affected by the strain rate change. The strain reversal also resulted in an increased cell size compared to that in monotonic straining. The misorientation distribution was also affected by the reversal, the number of lowest angle (the misorientation below 5°) boundaries being increased.

Finally, the transient behaviour of the flow stress in the ferrite following the abrupt change in the strain rate was modelled successfully in the ferrite using the Bergström's model, although several assumptions had to be incorporated.

A6.1 Introduction

Traditionally, deformation resistance and recrystallisation kinetics are determined in tests carried out at constant strain rate and temperature. Equivalent or average values of these external variables are used in modelling.

Due to chilling effect of the rolls, temperature at the surface layer of a slab drops fast during the rolling pass but it increases again immediately after the exit due to the conduction of heat towards the surfaces from the hotter core. Fast temperature changes are possible to simulate and perform on a Gleeble apparatus due to the use of resistive heating of the specimen in the experiment. In the previous ECSC project [A6.1], tests on a 0.07C-0.96Mn-0.046Nb-0.011Ti (wt. %) steel indicated that the final temperature, i.e. the temperature at which the static recrystallisation occurred after the deformation, determined the softening kinetics. If the initial temperature was more than 50°C higher than the final temperature in the deformation, the driving force remained somewhat lower due to a faster dynamic recovery at the high temperature and consequently the recrystallisation process became retarded. However, the surface temperature decreases in the rolling gap, not increases, compared to that during the subsequent pass interval.

In hot rolling, in each stands the strain rate changes continuously, rises immediately to a maximum and as the material leaves the roll gap it falls to zero (see [A6.2]). When strain rate is changed during deformation, these changes seem to produce transients while the stress is not instantaneously defined by a mechanical equation of state. In a rolling pass, strain is so short that stable conditions cannot be reached. Moreover, the strain required to achieve a new steady microstructural state is still higher [A6.1,A6.3-A6.5], and therefore the transients may have some influence on the recrystallisation kinetics subsequent to the deformation pass. These phenomena have been investigated in microalloyed steels earlier briefly by Karjalainen et al [A6.6,A6.7], but more results and firmer conclusions were searched in this project.

At the University of Oulu, experimental work was carried out aiming at investigating the influence of deformation history, i.e. the changes in temperature and strain rate in the course of deformation, on the flow stress and subsequent softening behaviour in the hot-deformed austenite and also in warm-deformed ferrite phase, and compare them with the behaviour of austenite investigated earlier at the University of Oulu and Corus (in compression) and also at CEIT (in torsion).

The effect of reversing straining was preliminarily investigated in the previous ECSC project in axial deformation and also in torsion [A6.1]. Some results have been published from axial compression tests in Refs. [A6.7,A6.8] and from torsion tests in Refs. [A6.9-A6.11]. It has been noticed that the reversal causes a distinct Baushinger effect in the flow stress and retardation in the static recrystallisation kinetics, which is dependent on the ratio of strains preceding and following the reversal [A6.8,A6.9]. Dynamic recrystallisation is delayed by the reversal [A6.9].

In the present study further information was aimed at clarifying the influence of strain reversal on the flow stress and the static recrystallisation kinetics of C-1.8Mn-0.13Ti-0.067Nb and 42MnV7 steels. However, it was realised important to fix the experimental conditions even better than earlier, and therefore a relatively high number of tests were first carried out using the same Nb steel (0.15C-1.42Mn-0.033Nb) as in the previous project [A6.1].

Furthermore, the aim was to investigate the effect of decreasing strain rate after the strain reversal on the recrystallisation kinetics, this way simulating the real deformation conditions present in the rolling gap. To modify the dislocation structure after the first pass deformation, in a few tests, a holding time resulting in partial recrystallisation was applied before the reversed deformation.

At the University of Oulu, computer modelling was not intended to perform in the project, only experiments to generate data and to improve the knowledge on the extent of the history effects and their importance to be incorporated in the models by the other partners. However, simple constitutive modelling on the influence of an abrupt change in the strain rate was performed based on the Bergström's model.

Before the experimental investigations of the constitutive behaviour, the evaluation of the existing constitutive models was performed by all partners. In fact, at the University of Oulu, the present research group uses regression models only, developed by partly the group itself and partly taken from the open literature. All regression models are based on average values of the deformation parameters. These models have been incorporated into some neural models to form an online hybrid model, installed at the moment in a steel company.

For the benchmark simulation, some calculations were carried out using the above-mentioned regression models for the static recrystallisation in order to adjust the interpass times applied in hot rolling tests at TU Freiberg.

A6.2 Experimental

A6.2.1 Experimental of temperature change tests

Two steels, a high-Ti bearing HSLA steel (marked here hTi-Nb) and a medium carbon steel 42MnV7, were used in the tests. Their chemical compositions are given in Table A6.1.

The cylinders 10 x 12 mm were machined from the pre-rolled plates for compression testing on a Gleeble 1500 thermo-mechanical simulator. The intention was to impose somewhat larger temperature changes than in the previous ECSC project [A6.1] and to investigate whether the chemical composition has any influence. Only the effect of abruptly reduced temperature, as the real case in rolling, was tested. Experimentally, a reasonable heating rate is 50°C/s so that the change of 100°C requires 2 s. The strain of 0.2 can be achieved within this time at the strain rate of 0.1 s⁻¹.

The test details are given in Table A6.2. To ascertain that the final temperature has been reached accurately before the beginning of recrystallisation, special schedules were adopted in some cases. In these tests, the temperature change was already accomplished shortly before the end of compression (e.g. strain of 0.16 occurred in the heating stage and strain of 0.04 at the final temperature, i.e. the time of 0.4 s, long enough, was available to adjust the temperature).

The lower strain rate of 0.01 s^{-1} executed in some tests provided a longer time, while the initial deformation could be carried out isothermally at the lower temperature, followed by heating in the final deformation stage to the final temperature (e.g. schedules such as 0.15 strain at 900°C followed by 0.04 strain during heating at 50°C/s to 1100°C or 0.16 strain at 1000°C followed by 0.02 strain during heating at 50°C/s to 1100°C).

The final (i.e. recrystallisation) temperature selected was 1100°C (a few tests were performed at 1150°C and some isothermal tests at temperatures between 1025°C and 1150°C to confirm the kinetics) providing a convenient recrystallisation rate and absence of any interfering precipitation even in the hTi-Nb steel. The lowest start temperature was 900°C , i.e. the change was then 200°C .

The stress relaxation technique was applied to determine the recrystallisation rate. The method has been established in many previous studies (see Ref. [A6.12], for instance).

A6.2.2 Experimental of strain rate change tests on the austenite

The strain rate changes taking place in the course of a rolling pass were simulated on a Gleeble 1500 simulator by three ways:

1. Starting at a low strain rate (0.01 s^{-1} in most cases) to a short strain of about 0.02 (stage 1), increasing instantaneously to a high strain rate (1 s^{-1} in most cases) to a certain strain (0.1 to 0.3) (stage 2), and finally decreasing instantaneously the strain rate back to a low value (stage 3).
2. Executing only the stages two and three (a high strain rate and the drop to a low strain rate) as described above.
3. The “continuous change” in the strain rate, one or two orders of magnitude, was created by changing the strain rate in small steps of 0.01 s^{-1} . The strain rate was first increased to a maximum at the first half of the applied strain and then decreased back to the initial level.

Two steels, hTi-Nb and a medium carbon steel 42MnV7, were used in these tests (see Table A6.1). The cylinders $10 \times 12 \text{ mm}$ were machined from the pre-rolled plates for compression testing on a Gleeble 1500 thermo-mechanical simulator. The stress relaxation technique was applied to determine the recrystallisation rate.

A6.2.3 Experimental of strain reversal tests on the austenite

All tests were carried out on a Gleeble 1500 simulator. The rods of 120 mm length and 10 mm in diameter, but the middle portion of 10 mm long thinned to the 9 mm diameter, were used. The thinned portion was essential to obtain a more uniform temperature zone where the deformation was concentrated. In this specimen, the gage length of 10 mm could be assumed, even though it appeared that this is not a very accurate estimate. It was found that the strain in the tension mode in the thinned portion became higher than nominal strain intended and after compression considerable bulging was present, so that the actual strain values had to be estimated from the diametric strains. The stress relaxation technique was employed to determine the recrystallisation rate. Due to experimental reasons revealed in the previous project [A6.1], in all tests the strain was tensile first following by a compressive strain in the second stage.

Three steels, Nb-steel, hTi-Nb steel and 42MnV7 were used in the tests. Various details were in the tests schedules as follows:

Ordinary T/C reversal tests

In all tests the first strain applied was constant, nominally 0.175, but the actual strain was about 0.2. The second strain was varied from the lowest value 0.0125, still controllable in the Gleeble, up to 0.4.

In addition, the austenite grain size was measured in the Nb steel after the reversed deformation varying the second strain between 0.05 and 0.125 in the tests.

T - C + decreasing strain rate

A decreasing strain rate from 1 s^{-1} (or 0.1 s^{-1}) to 0.01 s^{-1} was applied in the final stages (0.01 strain) of certain tests after the first tensile strain, followed by the equal compressive strain (nominally 0.175T/0.175C).

T-holding-C reversal tests

The holding stage of 3 s or 6 s after the first tensile strain at 1050°C was used in some tests to obtain a partially recrystallised structure before the straining in the compressive mode and finally followed by the stress relaxation.

A6.2.4 Experimental of strain rate change tests on the ferrite

The strain rate changes occurring in the course of a rolling pass were simulated on a Gleeble 1500 simulator by starting at a high strain rate (1 s^{-1} in most cases) to a certain strain (from 0.1 to 0.6), and finally decreasing instantaneously the strain rate back to a low value.

In order to reveal the possible influence of chemical composition, four steels, low-carbon, hTi-Nb, Ti IF and Nb-Ti IF, were used in the tests. Their chemical compositions (wt. %) are given in Table A6.1. The cylinders 10 mm x 12 mm were machined from the pre-rolled plates for Nb-Ti IF and low carbon steels, but smaller cylinders of 6 mm x 7.5 mm had to be used for Ti IF steel due to availability of the material. For IF steels, two pre-passes with the 10 min holding at 1000°C were applied to refine the grain size of austenite by recrystallisation and to enhance the recrystallisation rate in ferrite, which is known to be quite slow [A6.13]. This procedure was also adopted in a previous ECSC project [A6.1]. In the low carbon steel, a completely homogeneous polygonal ferrite structure was obtained by the cooling at the rate of 5°C/s to 500°C and then rising the temperature to the deformation temperature (700°C in most tests).

The stress relaxation technique was first tried to apply to determine the recrystallisation rate, but its sensitivity was found to be quite poor so that in most cases double-compression tests had to be employed, instead. The 5% total strain procedure was adopted [A6.12], in which the maximum stress (σ_3) is the final stress after the strain rate change, i.e., at the strain 0.02 in most tests.

Optical microscopy was used to examine the grain structures in different stages of the softening. Comparisons were carried out between the structures after constant strain rate straining and changed-strain rate deformation to reveal differences in the recrystallisation rate.

Electron microscopic techniques were employed to study the dislocation structures. Comparisons were carried out between the structures after constant strain rate straining and changed-strain rate deformation to reveal differences in the dislocation density or distribution. The deformed specimens were sectioned parallel to the compression axis and

the cut surfaces were examined metallographically using standard procedures. Specimens for thin foils were prepared by the twin-jet electropolishing in the solution of perchloric acid and ethanol at about -35°C. The foils were studied in a Leo 912 transmission electron microscope. Bulk specimens were examined in the SEM (Jeol 6400) using the backscatter electron image contrast. The misorientation between grains and subgrains were further analysed in the SEM using the Nordif CD100/Oxford INCA EBSD-unit attached.

A6.2.5 Experimental of strain reversal tests on the ferrite

The Armco iron was used to test the effect of strain reversal on the recrystallisation kinetics of ferrite. Stress relaxation technique was first tried, but the sensitivity was observed to be relatively poor so that the double compression technique (the 0.5% total strain [A6.12]) was employed instead. The first applied strain, always in tension, was kept constant, nominally about 0.18, but due to inhomogeneous deformation the diametric strain was higher, about 0.26. The second compression strain was varied from a small (≈ 0.02) value to about 0.7.

The specimens for dislocation studies were prepared similarly as described above in the case of the strain rate change tests.

A6.3 Results

A6.3.1 Results of temperature change tests

The influence of the temperature change was investigated for the austenite phase only. Figure A6.1 shows examples of the compression stress-strain curves for the hTi-Nb and 42MnV7 steels obtained in isothermal tests and in tests with increasing temperature. As starting at a lower temperature the stress increases faster than in the isothermal test. However, finally the curves approach each other, so that the difference between the final flow stress levels is generally relatively small (see values in Table A6.2), although the flow stress of the isothermal tests tended to remain lower. The hTi-Nb steel is an exception, for as starting the deformation from 900°C, a pronounced difference remains between the final flow stress values (Fig. A6.1(b)).

Typical relaxation curves recorded are shown in Fig. A6.2. From the curves the fraction v time data were computed and the Avrami-type equation fitted with the data. The complete data from the relaxation tests are given in Table A6.2, listing the stress in the beginning of relaxation, the Avrami exponent and the values t_5 , t_{50} and t_{95} for 5, 50 and 95% recrystallisation fraction, respectively.

The relaxation curves obtained for the hTi-Nb steel, with the start temperature of 900°C, are shown in Fig. A6.2(b). It is obvious from the straight relaxation lines that no recrystallisation takes place at all at 1100°C in this steel, even though recrystallisation is fast in the isothermal test at this temperature.

A6.3.2 Discussion on the results of temperature change tests

Recrystallisation behaviour is dominated by the total stored energy and its distribution (see e.g. Refs. [A6.14] and [A6.6]). The number of viable recrystallisation nucleation sites per unit volume (N_v) is dependent on local differences in the stored energy, whereas the growth rate is determined by the mean stored energy per unit volume (P_D). Consequently, the kinetics is affected by both of them, approximately as:

$$t_{50} \approx C P_D^{-1} N_v^{-1/3} \approx C P_D^{-1} [p \lambda S_v \delta^2]^{-1/3} \quad \dots \text{(A6.1)}$$

where C is a constant (related to the mobility of grain boundary) and p is a coefficient related to the probability of finding a sufficiently large nucleus (in fact, the probability of a nucleus being formed depends primarily on the standard deviation of the subgrain size distribution and on the mean subboundary misorientation (θ), which is why probability decreases rapidly as θ decreases), λ a geometrical factor, S_v the grain boundary area per unit volume (dependent on strain) and δ the subgrain size. P_D consists of influences of the dislocation density ρ_i , cell size δ and also misorientation θ as follows:

$$P_D = (Gb/10)\{\rho_i [1 - \ln(10b\rho_i^{1/2})]\} + (2\theta/b\delta)[1 + \ln(\theta_c/\theta)] \quad \dots (A6.2)$$

where θ_c is the critical angle for distinguishing between a grain and subgrain boundary ($\approx 15^\circ$), G the shear modulus and b is the burgers vector.

Hence, the recrystallisation kinetics is dependent on the stored dislocation density. At a lower temperature, at a given strain a higher stored energy in the form of numerous dislocations and vacancies can be expected owing to lesser dynamic recovery, enhancing the growth rate in the recrystallisation. Possible changes in the number and character of the nucleation sites are difficult to conclude.

The flow stress σ is also dependent on dislocation density as follows (see e.g. Refs. [A6.15] and [A6.8]):

$$\sigma = \sigma_i + \alpha_1 M G b \sqrt{\rho_i} + \alpha_2 M G b / \delta \quad \dots (A6.3)$$

where σ_i is the friction stress dominated by the viscous drag of solute atoms segregated to the moving dislocations, α_1 and α_2 are constants, M is the Taylor factor. Therefore, the flow stress level should be a (rough) measure of the dislocation density and consequently, the driving force of recrystallisation.

As seen in Fig. A6.1, as starting compression in a lower temperature, the flow stress increases higher, which results from lower thermal activation for dislocation glide and perhaps from higher dislocation density. The flow stress level in the end of compression reflects the driving force achieved for the subsequent recrystallisation. It can be seen that the difference between the final flow stresses is quite small, and it also appeared that scatter was present. As a consequence of small difference in the driving force, only a small variation in the recrystallisation rate may be expected, as observed from the measurements (Table A6.2). Generally, the results indicated that the change in temperature in the course of deformation (i.e. previous temperature history) only affects minimally the subsequent softening behaviour, but the actual/interpass temperature determines the rate.

There are some other things to be noted, however. It is well known that the deformation conditions, often described by the Zener-Hollomon parameter, vary in the thickness direction. The real temperature tends to be lower at the surface layer. The deformation rate varies in a complicated way. There may also be variation in the final strain reached in the deformation pass resulting from different deformation resistance in the layers at different temperatures. All these factors may result in differences in the recrystallisation kinetics in the thickness direction.

It should be noticed that if low temperatures below about 950°C exist (see Fig. A6.2(b)), they may cause prompt dynamic or/and static precipitation to occur which consequently retards significantly the recrystallisation. However, in practice, the duration of a rolling pass (strain/strain rate) is order of $0.3/3 \approx 0.1$ s, i.e. very short, and the temperature rise is fast

after the exit of the roll gap, so that enough time for any profound precipitation may not be available.

A6.3.3 Results of strain rate change tests on the austenite

The tests details for the hTi-Nb steel are shown in Table A6.3. Figure A6.3 shows two examples of the strain rate histories applied by the first way and the corresponding flow stress curves are given in Fig. A6.4 (also at certain constant strain rates for comparison). Fractional softening curves are plotted in Fig. A6.5, following these strain rate changes or at constant strain rates.

Typical strain rate histories according to the second way are shown in Fig. A6.6 and the corresponding flow stress curves in Fig. A6.7 and the fractional recrystallisation curves in Fig. A6.8, the latter also containing curves at constant strain rate for comparison.

For medium carbon steel, the strain rate history is shown in Fig. A6.9(a) and the flow stress curves in Fig. A6.9(b). Stress relaxation curves and fractional recrystallisation curves are given in Fig. A6.10(a and b), respectively.

Immediately following an abrupt decrease in the strain rate, the control of the strain rate seems to be shortly poor, and the flow stress drops instantaneously to a low value but increases back to a proper level corresponding to the controlled strain rate.

It is obvious that the flow stress remains temporarily somewhat higher than the flow stress at a final strain rate in constant strain rate test (below or close to the level of the flow stress in a test at the constant average strain rate). Hence, the flow stress is slightly lagging as following the changes in the strain rate compared to that expected from the mechanical equation of state.

With increasing strain at the lower final strain rate, the flow stress gradually approaches the level of flow stress of a constant strain rate test. In the real rolling pass, the final strain is, however, extremely short (decrease of the strain rate so fast), so that it can be concluded that the flow stress at the exit is markedly higher than that based on the final strain rate (approaching zero).

In the instance of continuously changing strain rate, the final flow stress is practically independent of the previous strain rate history, as seen in Fig. A6.11(a). The recrystallisation rate is, however, dependent on the history so that the high strain rates applied results in faster recrystallisation kinetics than the lower ones, as seen in Fig. A6.11(b).

The previous results [A6.8] showed that the MFS of the deformation (strain 0.3) is about 12% higher under the varying strain rate than under a constant strain rate equal to the mean strain rate. The values of MFS were 87 MPa (at the changing $\dot{\epsilon}$) and 78 MPa (at 0.064 s^{-1}). This may mean that the real rolling load tends to be higher than the value predicted from constant strain rate tests.

Recrystallisation rate (in the case of a short final strain, 0.005-0.02 strain) is close to the value at the high strain rate applied, i.e. it lags clearly more than the flow stress (Figs. A6.5 and A6.8). Also, as shown earlier [A6.7,A6.8], the average strain rate can be used to model the static recrystallisation kinetics reasonably.

At higher strains, and for 42MnV7 steel in particular, it seems that following the decrease in the strain rate, dynamic recrystallisation will be initiated which results in the decreasing rate of metadynamic recrystallisation, and even absence of recrystallisation in the case of 0.01 strain as seen in Fig. A6.10.

Both steels behave qualitatively in a similar way so that no significant effects of the chemical composition can be found. The peak strain of 42MnV7 steel is shorter than that of hTi-Nb steel. The results confirm the observations for Nb-bearing steel obtained earlier [A6.7,A6.8]. Some discussion concerning the role of the internal state variables can be found in some previous papers [A6.7,A6.8].

A6.3.4 Results of strain reversal tests on the austenite

The main interest was paid on the kinetics of SRX after a very short second strain following the reversal from tension to compression. Different results have been published regarding this behaviour. From axial compression tests results showed that the SRX kinetics is not affected by the short second strain and higher second strains are enhancing the rate, although the rate is always smaller than in monotonic deformation at a comparable total strain [A6.6]. In torsion tests, a retardation (a minimum) of the kinetics was observed at small second strain, i.e., the rate was slower than the rate without any second strain (here in the CEIT contribution) [A6.9,A6.10,A6.11]. At higher second strains the rate increased but remained lower than in monotonic deformation at a comparable total strain.

The main observations are as follows:

Tension - compression tests

The Bauschinger effect of about 5% is present in all the steels tested as shown by the flow stress curves in Fig. A6.12 in four examples.

As seen in Fig. A6.12, the flow stress in tension measured using the long rods is higher than that in compression determined using cylinders. This is probably due to differences in the distribution of deformation (centrally thinned specimen; shoulders) or in the prediction of the flow stress (problems to define the exact gauge length; non-uniform temperature).

The diametric strain (C-strain) is higher than the longitudinal strain (L-strain) indicating some barrelling occurring in the middle zone. Therefore, the actual strains before and after the reversal are higher than intended (nominal) ones. Control of small strain less than about 0.015 is difficult, for the proper stress level has not been reached at the 0.0125 strain, as seen in Fig. A6.13.

The time for 50% recrystallised fraction t_{50} is shown in Fig. A6.14, as a function of the second strain following the tensile strain of 0.175 (the actual C-strain about 0.2, calculated assuming the gage length of 10 mm). For all the steels tested, the trend in t_{50} is the same, first increasing with increasing second strain up to a maximum at about 0.04 strain (for Nb-hTi at about 0.02), which is smaller than 0.1 in torsion [A6.10,A6.11], and then decreasing towards the value determined for the metadynamic recrystallisation in the monotonic (i.e., without reversal) deformation, reached at about 0.5 strain in the case of Nb-steel (Fig. A6.14(a)). Previous data (nominal strains) fit the present data for the Nb-steel, but earlier only higher second strains were used. In the present tests, distinct reduction in the SRX rate is observed at small second strains, in agreement with the torsion test results from CEIT [A6.10,A6.11], such a behaviour which was not detected in the previous tests.

For the Nb-bearing steel, the maximum t_{50} (i.e. the minimum rate) is 3 times longer than that in monotonic deformation at 0.2 strain without any second strain (about 20 s and 6.2 s, respectively). The latter value of t_{50} is reached again at the 0.14 second strain, i.e., at the strain somewhat smaller than the first strain. The MDRX behaviour was reached at about 0.25 strain in the monotonic mode, and now in the reversed deformation at about 0.5, i.e. at the total strain of 0.7, or, in other words, the second strain of $0.5 - 0.14 = 0.36$ is needed to enhance the SRX rate from the monotonic t_{50} value of 6 s to MDRX value of about 3 s. The MDRX rate is not affected by the reversal, as observed earlier [A6.8].

After the maximum, the slope $\log(t_{50})$ v $\log$ (time) curve is much lower than that in the monotonic deformation, i.e. the power of strain is lower in the reversed deformation than in the monotonic deformation (about -3 [A6.8]).

Retardation of the SRX rate can be explained by some kind of dissolution of the deformation dislocation structure due to reversed direction of straining [A6.8-A6.11].

For the hTi-Nb steel, the retardation is somewhat smaller, about 2 times (Fig. A6.14(b and c)), for 42MnV7 steel about 3 times. Hence, the retardation is always of the same order for all these steels.

Tension - compression - decrease in strain rate

A few tests were carried out for Nb and 42MnV7 steels using reversed deformation 0.175T/0.175C and decreasing the strain rate from 0.1 s⁻¹ to 0.01 s⁻¹ in the end of second deformation for short 0.01 strain, before relaxation to determine the SRX rate (see Table A6.3). Figure A6.15 shows the flow stress behaviour and the relaxation curve recorded. The relaxation curves were of the proper shape so that the schedule is experimentally well under control. The data obtained are included in Fig. A6.14 for these steels revealing a slight retardation compared to the kinetics without decreasing the strain rate, but the effect is quite small. The effect of first/second strain ratio was not investigated.

Tension - holding - compression

A few tests were carried out to investigate the influence of holding during reversal on the subsequent SRX rate. During holding some SRX can occur and presumably rearrangements in the deformation dislocation structure. Medium carbon steel 42MnV7 was used in the tests. Two different holding times were selected, 3 and 6 seconds resulting in about 30% and 75% recrystallised fractions, respectively. The flow stress curves (L and C-strain) as holding for 6 s are shown in Fig. A6.16 and the fraction v times curves obtained from the relaxation curves for two cases: without holding and with 6 s holding. From the fraction curves it can be noticed that the holding has slightly increased the softening rate, which might be explained as a result of elimination of first deformation effects in holding by recovery and recrystallisation so that the second strain (or the total strain) becomes more effective. Contrary to this behaviour, in the monotonic deformation, partial recrystallisation occurring between the passes would decrease the kinetics. The data are also inserted in Fig. A6.14. The second feature observable in the figure is that the final part of the fraction data falls well below the Avrami curve, even though predicted using a small Avrami exponent of 1.2. This might reflect the influence of those grains, which recrystallised during holding and have no retained deformation. As seen, the fraction curve for 0.175 strain without reversal is much steeper.

A6.3.5 Results of strain rate change tests on the ferrite

The test details for IF steels and hTi-Nb steel are given in Table A6.4 and Table A6.5, respectively, and the tests details and main results for the low-C steel are listed in Table A6.6. Two examples of flow stress curves in tests carried out at 800°C at constant high (1 s^{-1}) and low (0.01 s^{-1}) strain rates and changing the strain rate in the later part of the test (the total strain 0.3) are plotted in Fig. A6.17. It is clear that for both Ti IF and Nb-Ti IF steels, the final stress level in the strain rate change test remains above the flow stress at the lower strain rate, but below the level of the flow stress at the initial high strain rate. Qualitatively this is the same result as observed in the case of hot-deformed austenite.

Figure A6.18 gives examples of the recorded relaxation curves in the Ti IF steel, Fig. A6.18(a) for the austenite (at 1000°C) and Fig. A6.18(b) for the ferrite phase (at 800°C). It is obvious that the shape of the relaxation curve is much flatter for ferrite than for austenite, so that it is inaccurate to determine the recrystallisation start and finish times for the ferrite. Nevertheless, some data and the Avrami type curves fitted with the data for the recrystallised fraction v time are plotted in Fig. A6.19 and in Fig. A6.20. In the former figure, the data and curves are given for both austenite and ferrite to reveal the differences between them. They show that the recrystallisation kinetics in ferrite occurs over a longer time, i.e. the Avrami exponent is much lower than that for the austenite, 0.5 and 1.4 for the ferrite and austenite phases, respectively. It was observed in Ref. [A6.13] that the softening in deformed ferrite takes place mainly by the recovery process. It is known that the recovery decreases the Avrami exponent so that probably some influence of recovery is included in the behaviour of ferrite, even though usually the recovery effects can be excluded by the stress relaxation analysis [A6.12, A6.16]. In Fig. A6.20 the data and curves from constant strain rate tests and from strain rate change tests are plotted, but distinct differences in the kinetics cannot be distinguished due to considerable scatter in the data. On the other hand, it can be concluded that the differences cannot be very pronounced.

Typical relaxation curves for the Nb-Ti IF steel are displayed from two monotonic tests and from one strain rate-change test at 800°C in Fig. A6.21. However, one can notice that the recrystallisation of ferrite cannot be seen in the relaxation curves. Otherwise, the shapes of the curves are quite similar.

The kinetics of recrystallisation was expected to be relatively fast in low-carbon grades so that one of them was selected as a material in the further tests (Table A6.6). However, the relaxation curves were insensitive to reveal recrystallisation so that the double-compression technique was adopted to determine the softening kinetics (the 5% total strain procedure [A6.12] in a slightly modified way).

Typical behaviour of flow stress is displayed in Fig. A6.22 in two cases, the holding time 250 s or 400 s before the second pass (which does not affect the flow stress of the first pass, of course). The second strain following the change in the strain rate was only 0.02, so that the flow stress curve is short. It can, however, be seen that the flow stress is much below the level at the high strain rate, and in fact, closer to the flow stress at the lower strain rate, but anyhow lagging behind a mechanical equation of state.

In Fig. A6.23 data for the recrystallisation are plotted, after the monotonic deformation of 0.6 strain at 1 s^{-1} or 0.01 s^{-1} , and after the strain rate change tests at 0.58 strain. The considerable amount of the data is to reveal that following the strain-rate-change deformation, the kinetics is retarded considerably and intermediate to those after monotonic deformations. However, the recrystallisation seems to be completed after the strain-rate-change deformation similarly as at 1 s^{-1} , which is contrary to that at 0.01 s^{-1} , when the kinetics is extremely slow and only small recrystallised fractions are achieved. Barnett et al [A6.17] have found the power of -0.15 for the strain rate, which does not fit with the

present data, but Jansen et al [A6.13] reported that the complete recrystallisation is only achieved at high strains and high temperatures in ferrite.

Some relaxation tests were tried using Armco iron deformed at 800°C ($0.28 \text{ at } 1 \text{ s}^{-1} + 0.02 \text{ at } 0.01 \text{ s}^{-1}$), expecting a faster kinetics. However, the relaxation curves were quite flat, so that the determination of recrystallisation kinetics seemed to be difficult.

Typical partially recrystallised grain structures in low carbon steel are shown in Figs. A6.24-A6.27. These microstructures confirm nicely the results in Fig. A6.23 revealing that the softening is retarded considerably. Fully recrystallised grains are uniformly white in colour while the others are darker and their surfaces seem to be wrinkled due to the internal cell structure. Whether those grains are only deformed (dynamic recovery) or recovered after the deformation (static recovery) is more difficult to distinguish.

Examples of the dislocation structures in the Ti-IF steel after monotonous deformation and after the changed-strain rate deformation are shown in Fig. A6.28 and Fig. A6.29, respectively. It can be seen that in both cases the cell structure has formed. Measuring the diameter of about 100 cells in the constant strain rate test and the diameter of 425 cells in the changed-strain rate test the mean cell sizes were $1.4 \text{ }\mu\text{m}$ and $1.8 \text{ }\mu\text{m}$, respectively. The values are given in Table A6.7. This seems to indicate that somewhat larger cells are present after changing the strain rate than at the constant (high) strain rate.

It appeared that the cell structure could be revealed on the surface of the bulk polished specimens by the SEM-BEI contrast that enabled much easier measurements of the sizes of more numerous cells. Figure A6.30 and Fig. A6.31 show examples of the cell structures. Analysis of more than 2000 cells gave the values for the diameter as $2.2 \text{ }\mu\text{m}$ and $2.8 \text{ }\mu\text{m}$ for the constant and the changed strain rate deformation conditions, respectively. Once more, the cells seem to be larger after the changed the strain rate, in agreement with TEM observations. The cell size determined in SEM-BEI is, however, much larger than that from TEM.

The distributions of misorientation of the grain boundaries were determined by SEM-EBSD, and the results are shown in Fig. A6.32. It can be noticed that the change in the strain rate does not affect the misorientation distribution, neither in the high-angle nor low-angle range. The same was observed for the low-C steel, too.

A6.3.6 Results of strain reversal tests on the ferrite

Typical flow stress curves in the strain-reversed deformation are plotted in Fig. A6.33. A pronounced Bauschinger effect, about 12%, can be seen, i.e. higher than in the austenite. Figure A6.34 shows the recrystallised fraction in 40 s, during the holding time that was held constant in the first set of tests. It is seen that the recrystallised fraction remains distinctly smaller than without any second strain if the second strain is low, 0.02 - 0.15, i.e. a small reversed strain even retards the recrystallisation kinetics of the ferrite, similarly as observed earlier for the austenite. The time t_{50} increases from 20 s to 200 s, as shown in Fig. A6.35. These values are computed by fitting the Avrami type curve with the exponent of 0.9 with the measured fraction at 40 s. Figures A6.34(b) and A6.35(b) show the fractions after monotonic and reversed deformations as a function of absolute strain, i.e. the first tensile+second compressive strains. The maximum retardation can be estimated from Fig. A6.35(b), and it is about 8 times at the total strain of 0.35. It is also seen that there is no single power of strain in a reversed deformation, while in the monotonic deformation the power is -1.8 . This value is somewhat lower than generally for austenite or reported for ferrite [A6.4].

Recrystallisation was also followed by metallographic means and the retardation could be detected in optical microstructures. Another observation is that at small/intermediate reversed strains, not only the kinetics are retarded but also the recrystallised grain size tends to be much coarser, and particularly a few coarse ($> 500 \mu\text{m}$) grains appear in the partially recrystallised microstructure (Fig. A6.36). In recrystallisation (within 20 s) after compression to the strain of 0.20 at 1 s^{-1} , the grain size became slightly coarser than the initial grain size (Fig. A6.37(a and b)). This is in consistence with the equation proposed by Barnett et al [A6.17] for the recrystallised grain size in the low-C ferrite. However, after small reversed strain, the grain size is much coarser, but it becomes refined with increasing total strain, as shown in the recrystallised microstructures (holding just until the complete recrystallisation (about 85% fraction) in each case), compare Fig. A6.37(b-d), for instance. The evolution of the grain size is plotted in Fig. A6.38. In an Al-Mg alloy, Zhu and Sellars [A6.18] observed the grain size to become much coarser in the reversed straining with the increasing second strain.

The dislocation structures seen in TEM in Armco iron after compression at 1 s^{-1} to the 0.3 strain and after straining 0.23 in tension + 0.07 in compression are shown in Fig. A6.39 and Fig. A6.40 respectively. In the both cases, cell structure is present. The analysis of about 150 cells resulted in the cell size of $2.8 \mu\text{m}$ and $4.1 \mu\text{m}$ without the strain reversal and with the strain reversal, respectively (see Table A6.7). Hence, the reversed deformation seems to lead to distinctly larger cells.

The cell structures revealed by SEM-BEI are shown in Fig. A6.41 and Fig. A6.42 for the deformation without and with the strain reversal, respectively. Surprisingly, measuring the size of 2000-2700 cells resulted in the identical size of $4.5 \mu\text{m}$, for the both deformation histories.

The distributions of misorientation of the grain boundaries are shown in Fig. A6.43 revealing that there are more boundaries with a low-angle misorientation in the instance of reversed deformation than without the reversal. Hence, reversing the direction of straining tends to decrease the average misorientation between cells, i.e. increase the number of lowest-angle boundaries (misorientation below 5°) at the expense of boundaries with the $6\text{-}8^\circ$ misorientation.

A6. 4 Strain rate change tests for constitutive modelling at CSM

A large number of strain rate tests on 55Cr3 steel were carried out and in cooperation with CSM, and their analysis was performed to include in the constitutive modelling (see the CSM report).

A6. 5 Modelling of the transient behaviour after the change in the strain rate

The transient observed in the flow stress in the ferrite (in austenite as well) following the abrupt drop of the strain rate was modelled in order to try to understand the factors controlling the evolution of the flow stress, lagging behind the mechanical equation of the state, as observed earlier. The Bergström's model, a relatively simple approach used at MEFOS, for instance, to model the flow stress behaviour under constant deformation conditions was adopted. In the model, the flow stress is related to the total dislocation density and affected by the rates of dislocation immobilisation and remobilisation (recovery). The parameters have to be estimated under the transient conditions, changing exponentially after the drop of the strain rate.

A number of physical material models have been developed by different research groups describing the evolution of flow stress during deformation, which essentially include the material dislocation density as a history parameter. Of the various models, the Bergström's model [A6.19] for plastic deformation of metals is relatively simple, and it has been explored under constant deformation conditions in the present and the previous ECSC project by MEFOS, so that this model has been adopted in this study to model the effects of the abrupt change in the strain rate on the observed flow stress transient.

Recently Bianchi and Karjalainen [A6.20] have incorporated the change in the strain rate in the model for dynamic and metadynamic recrystallisation in rod rolling.

A6.5.1 Bergström's model

The true flow stress, σ , of crystalline materials is related to the total dislocation density ρ_i as

$$\sigma = \sigma_{i0} + \alpha G b \sqrt{\rho_i} \quad \dots (A6.4)$$

where σ_{i0} is a strain-independent friction stress, α is a constant, G the shear modulus and b the dislocation burgers vector. The change in the dislocation density is

$$d\rho_i / d\varepsilon = U(\varepsilon) - \Omega \rho_i \quad \dots (A6.5)$$

where $U(\varepsilon)$ is the measure of the rate of dislocation immobilisation and Ω is a measure of dislocation remobilisation (also termed as the recovery parameter). The integration results in

$$\rho_i = U/\Omega (1 - \varepsilon^{-\Omega\varepsilon}) + \rho_0 \varepsilon^{-\Omega\varepsilon} \quad \dots (A6.6)$$

where ρ_0 is the "grown-in" dislocation density, in the well-annealed materials order of $2 \times 10^{12} \text{ m}^{-2}$. Combining Equation (A6.6) with Equation (A6.4), the following σ - ε relationship is obtained.

$$\sigma = \sigma_{i0} + \alpha G b [U/\Omega (1 - \varepsilon^{-\Omega\varepsilon}) + \rho_0 \varepsilon^{-\Omega\varepsilon}]^{1/2} \quad \dots (A6.7)$$

The experimental and the fitted curves using Equation (A6.7) are plotted in Fig. A6.44. The values of the parameters obtained are listed in Table A.6.8.

It is assumed that Ω consists of an athermal component, Ω_0 , and a thermal component $\Omega(\varepsilon', T)$, i.e.,

$$\Omega = \Omega_0 + \Omega(\varepsilon', T) \quad \dots (A6.8)$$

$$\Omega(\varepsilon', T) = \{k n_0 (2D_0)^{1/2}\}^{2/3} \exp[-Q_m/(3RT)] \varepsilon'^{-1/3} \quad \dots (A6.9)$$

where k is a constant and n_0 is the number of vacancies per unit volume, D_0 the diffusion coefficient and Q_m the activation energy of grain boundary migration. For ferrite, using D_0 value of $2 \text{ cm}^2/\text{s}$ and considering $Q(\text{ferrite}) = 2xQ_m(\text{ferrite}) \approx 30000 \text{ cal/mol}$, and kn_0 value of 1.85×10^{-3} was obtained by Bergström. Ω in the present constant strain rate condition, obtained from fitting, is shown in Fig. A6.45. In the following, the modelling of transient behaviour is described.

A6.5.2 Estimation of Ω in the very beginning of the transient

According to the definition of the recovery parameter, Ω , it is a measure of the probability the immobile dislocations to remobilise. The athermal component Ω_0 was determined experimentally to be ≈ 10.8 under the constant strain rate conditions (see Fig. A6.45). The value for $\Omega(\varepsilon', T)$ can be calculated from Equation (A6.6) as function of temperature and strain rate.

However, the athermal component, Ω_0 , obviously gets enhanced to a new value of Ω_{SRC} at the very moment of the strain rate change owing to the shift of a high dislocation density to the lower strain rate condition, amenable for remobilization and recovery. We assume that the enhancement is relative to the ratio of dislocation densities, i.e.

$$\Omega_{\text{SRC}} = (\rho_i \text{ at first } \varepsilon' / \rho_i \text{ at second } \varepsilon') * \Omega_0 \quad \dots \text{ (A6.10)}$$

Finally the total Ω in the start of the transient is

$$\Omega_{\text{initial}} = \Omega_{\text{SRC}} + \Omega(\varepsilon', T) \quad \dots \text{ (A6.11)}$$

A6.5.3 Ω during the transient

Now the value of Ω_{initial} is known and the final value of Ω is that at the constant second (final) strain rate. However, to model the behaviour of Ω during the transient, the duration of the transient must be known.

We can assume that the ratio of dislocation densities affects the time before the remobilization of immobile dislocations. According to Bergström, the average rest time the immobilized dislocations remain in rest before remobilization is

$$\tau = 1/(\Omega \varepsilon') \quad \dots \text{ (A6.12)}$$

Hence, under the transient

$$\tau_{\text{tr}} = (\rho_i \text{ at first } \varepsilon / \rho_i \text{ at second } \varepsilon) * \tau \quad \dots \text{ (A6.13)}$$

For instance, for the change from 1 s^{-1} to 0.01 s^{-1} τ is 18.8 s.

The length of the transient strain is

$$\varepsilon_{\text{trans}} = (\rho_i \text{ at first } \varepsilon / \rho_i \text{ at second } \varepsilon) * \tau * \varepsilon' \quad \dots \text{ (A6.14)}$$

Under *constant* strain rate conditions, τ at 1 and 0.01 s^{-1} are 0.087 s and 5.55 s, respectively (Ω at 1 and 0.01 s^{-1} are 11.5 and 18, respectively, from the fitted curves). See Table A6.8 for these values and the estimated transient strain.

Therefore, at the start of the transient, the time spent to mobilise dislocations for the recovery at $0.01 \text{ s}^{-1} \Rightarrow (\rho_i \text{ at } 1 \text{ s}^{-1} / \rho_i \text{ at } 0.01 \text{ s}^{-1}) * 5.55 \text{ s} \approx 18 \text{ s}$. This means that, following the strain rate change, Ω changes from 42.2 to 18 in about 18 s (i.e., $\varepsilon = 0.18$ at 0.01 s^{-1}) to reach the steady-state value of 18 and it has been assumed that there is an exponential change in Ω in the transient region (fitted by regression).

Then an exponential equation was assumed to describe the course of Ω during the transient:

$$\Omega = \Omega_{\text{changed } \dot{\epsilon}'} + (\Omega_{\text{SRC}} - \Omega_{\text{changed } \dot{\epsilon}'}) [\exp(-k\epsilon_{\text{SRC}})] \quad \dots (A6.15)$$

where

$$\begin{aligned} \Omega_{\text{changed } \dot{\epsilon}'} &= \Omega \text{ at the changed (second) strain rate} \\ \Omega_{\text{SRC}} &= \Omega \text{ at the start of the transient} \\ k &= \text{a constant from the regression} \\ \epsilon_{\text{SRC}} &= \text{strain following the strain rate change.} \end{aligned}$$

The curves Ω v strain are plotted in Fig. A6.46.

A6.5.4 Estimation of the friction stress σ_{i0} at the start of the transient

The abrupt drop in the strain rate leads to a high dislocation density inherited from the first (high) strain rate deformation. Following the strain rate change and during further straining at a low strain rate, the excess dislocations in the subgrains lead to a very high value of ρ_m and the friction stress drops to a very low value before increasing back to its steady-state value during the transient strain.

An attempt has been made to roughly estimate the friction stress at the start of the transient as using the Equation (A6.3) given by Furu et al [A6.15] reproduced below and some rough mathematical factors. The equation for the flow stress of work hardened materials based on the state variables, the dislocation density ρ_i and the cell size δ is

$$\sigma = \sigma_{i0} + \alpha_1 M G b \sqrt{\rho_i} + \alpha_2 M G b / \delta$$

where α_1 and α_2 are constants and M is the Taylor factor. The friction stress σ_{i0} is dominated by the viscous drag of the solute atoms segregated to the moving dislocations and it can be expressed as

$$\sigma_{i0} = [\beta / (M D_0 b^2)] Z / \rho_m \quad \dots (A6.16)$$

where β is the drag force, D_0 the diffusion coefficient, Z the Zener-Hollomon parameter and ρ_m the mobile dislocation density in ρ_i .

Let us assume that β in Equation (A6.14) does not change significantly during deformation so that the factor $[\beta / (M D_0 b^2)] \exp(Q_{\text{def}}/RT)$ can be considered to be constant (Q_{def} is the activation energy of deformation).

Therefore, Equation (A6.16) can be rewritten as

$$\sigma_{i0} = K \epsilon' / \rho_m \quad \dots (A6.17)$$

where K is a constant equal to $[\beta / (M D_0 b^2)] \exp(Q_{\text{def}}/RT)$.

It was experimentally determined from the constant strain rate flow stress curves (Fig. A6.44) that σ_{i0} values are about 35, 29, 25 and 20 MPa at 1, 0.1, 0.01 and 0.001 s⁻¹, respectively.

The mobile dislocation density is not known, only the total one at the constant strain rate conditions. The relative values can, however, be written for the conditions 1 and 2

$$\rho_{m2} / \rho_{m1} = (\epsilon_2 / \epsilon_1) (\sigma_{i01} / \sigma_{i02}) \quad \dots (A6.18)$$

Using Equation (A6.18) and the friction stress values at four measured strain rates, it can be shown that

$$\rho_{m\ 1.0} = \rho_{m\ 0.1}/0.12 = \rho_{m\ 0.01}/0.014 = \rho_{m\ 0.001}/0.00175 \quad \dots (A6.19)$$

where the numbers in the suffixes represent the strain rate. Let the mobile dislocation density at $1\ s^{-1}$ be $(\rho_{m\ 1.0}) = x$. Then $\rho_{m\ 0.1} = 0.12x$, $\rho_{m\ 0.01} = 0.014x$ and $\rho_{m\ 0.001} = 0.00175x$.

A plot of $\varepsilon'/\rho_m * 1/x \ v \ \sigma_{i0}$ as using the data at constant strain rate conditions gives the following straight line relation (Fig. A6.47).

$$\varepsilon'/\rho_m = (1/\rho_{m1}) * 0.02864 \ \sigma_{i0} \quad \dots (A6.20)$$

Further, it is assumed here that U (the measure of the rate of dislocation immobilisation) does not change significantly as a consequence of abrupt strain rate change and the values assigned at the constant strain rate conditions are still valid.

It is assumed that ρ_m in the instance of strain rate change gets approximately multiplied by the ratios of the U and ρ_i at the two strain rates. Therefore,

$$\rho_{mSRC} = \rho_{mfinal} * (U_{first}/U_{final}) * (\rho_{i\ first}/\rho_{i\ final}) \quad \dots (A6.21)$$

For instance, at the start of the transient $\rho_{m0.01} = 0.014\rho_{m1} * (U_{1.0}/U_{0.01}) * (\rho_{i1.0}/\rho_{i0.01}) \approx 0.0949 \ \rho_m$.

From the plot $\varepsilon'/\rho_m * 1/x \ v \ \sigma_{i0}$, Fig. A6.47, the value of the friction stress can be obtained, about 3.7 MPa in the present example. This also suggests that the mobile dislocation density has approximately increased by nearly $0.0949x/0.014x = 6.8$ times owing to the drop in the strain rate by two orders of magnitude from 1 to $0.01\ s^{-1}$. Similarly, it can be shown that the friction stress values at the start of the transients at 0.1 and $0.001\ s^{-1}$ following 1 and 3 orders of magnitude drop in strain rate (from $1\ s^{-1}$), respectively will be about 13.8 MPa and 1.1 MPa. The corresponding increase in mobile dislocation densities will be about 1.9 and 19.2 times that at the constant strain rates, respectively.

A6.5.5 The behaviour of the friction stress during the transient

The course of the friction stress during the transient flow is not known. Therefore, it had to be obtained by the fitting with some experimental curves. This way, the equation describing the evolution of during the transient strain is

$$\sigma_{i0} = \sigma_{i0SRC} + [(\sigma_{i0\ changed\ \varepsilon'} - \sigma_{i0SRC})*(1-\exp(-C\varepsilon_{SRC}))] \quad \dots (A6.22)$$

where

- σ_{i0SRC} = friction stress at the start of the transient
- $\sigma_{i0\ changed\ \varepsilon'}$ = friction stress at the changed strain rate
- C = a constant from regression
- ε_{SRC} = strain following the strain rate change.

The curves are shown in Fig. A6.48.

Finally, all parameters needed in Equation (A6.7) for the flow stress are known, and the flow stress can be predicted at constant as well as transient conditions. The results are shown in Fig. A6.49.

A6.4 Conclusions

It is concluded that:

1. The final flow stress after compression under increasing temperature from 900 to 1100°C is only slightly higher than that after isothermal compression at 1100°C, i.e. the stored dislocation densities are almost equal. The chemical composition (microalloying) of the steel has no detectable influence on this behaviour.
2. The results indicate that the chilling effect of rolls in the course of deformation, i.e. the *previous temperature history*, only affects minimally the subsequent softening behaviour of the steel, but it is the *actual* interpass temperature which determines the softening rate.
3. In microalloyed steels, precipitation may occur at low testing temperatures preventing the subsequent recrystallisation. However, in rolling the surface temperature drops in the rolling gap are very short in duration so that precipitation can hardly occur to any significant extent.
4. Under abruptly decreasing strain rate, in the dynamic recovery regime the flow stress clearly lags behind the behaviour of a mechanical equation of state. After the drop, the final flow stress remains distinctly higher than that at the corresponding constant final strain rate, but it is close to or lower than the stress at the constant mean strain rate of the whole deformation history. Therefore, the mean flow stress is distinctly higher under the varying strain rate than at constant mean strain rate.
5. The static recrystallisation rate lags even more than the flow stress from the behaviour following the mechanical equation of state. The recrystallisation rate remains close to that at the high strain rate before the fast drop of the strain rate, if the strain at the low strain rate is short (0.005 to 0.02). However, because the effect of strain rate on the static recrystallisation kinetics is small, the average strain rate can be used to fair accuracy to model the static recrystallisation kinetics also after a strain rate change deformation.
6. If the strain rate drop takes place after a strain resulting in dynamic recrystallisation at the low strain rate, the subsequent metadynamic recrystallisation behaviour is affected significantly, the magnitude depending on the strain after the strain rate change. The recrystallisation may even be totally absent. The average strain rate is a poor variable for modelling this behaviour.
7. A special centrally thinned rods are used as specimens in the tests but it has been realised that reversal tests are difficult to control. Flow stress in tension seems to be higher than in compression of cylinders, obviously due to experimental difficulties to control the gage length as well as strain and temperature distributions along the rod.
8. Bauschinger effect can be seen in the flow stress in the connection with the strain reversal.

9. Strain reversal causes retardation in the SRX rate in all the three tests used in the study, the amount depending on the magnitude of the second strain. The minimum rate occurs at about 0.02-0.04 second strain (if the first actual strain is 0.2), being 2-3 times lower than the rate without any second strain. At second strains beyond 0.14, the SRX rate becomes faster than that without any second strain. This effect can be explained by partial dissolution of deformation dislocation structure after reversing the strain direction.
10. Changing the strain rate in the end of second strain results in a decreased SRX rate. Experimentally these effects are possible to investigate.
11. Partial recrystallisation after first strain before the reversed deformation seems to result in an enhanced SRX rate, even though the effect is relatively small.
12. Under abruptly decreasing strain rate, the flow stress of ferrite clearly lags behind the behaviour of a mechanical equation of state. After the drop, the final flow stress remains distinctly higher than that at the corresponding constant final strain rate. The behaviour is similar in Ti IF, Nb-Ti IF and low carbon steels and qualitatively identical with the behaviour of austenite of the steels.
13. The static recrystallisation rate of the ferrite lags considerably from the behaviour according to the mechanical equation of state. The kinetics at the strain shortly after the strain rate change (strain 0.02) is between those at initial (1 s^{-1}) and final strain rates (0.01 s^{-1}).
14. The softening of ferrite is affected by the recovery. The stress relaxation technique is not very sensitive to reveal the recrystallisation behaviour.
15. In the ferrite (Armco iron), the strain reversal results in a decreased SRX rate at small reversed strains, the maximum effect being about 8 times.
16. In the ferrite, the reversal results in a significantly coarsened grain size at small reversed strains.
17. Dislocation cell structures can be revealed in the ferrite by TEM and SEM-BEI. Grain boundary misorientation distributions can be determined by SEM-EBSD.
18. After changing the strain rate abruptly, the cell size seem to be larger than at constant high strain rate revealing that the driving force for the static recrystallisation decreases rapidly after the drop of the strain rate.
19. The misorientation distribution is not affected by the strain rate change.
20. The strain reversal also results in an increased cell size compared to that in monotonic straining.
21. The misorientation distribution is also affected by the reversal, the number of lowest angle (misorientation below 5°) boundaries being increased.
22. The Bergström's model can be used to model the transient in the flow stress in the ferrite following the drop in the strain rate, but several assumptions have to be made.

References

- A6.1 Zhou, M, et al: The effect of strain reversal and strain-time path on constitutive relationships for metal rolling/forming processes, Report EUR 19891 EN, 2001.
- A6.2 Sellars, C M: Recrystallization and Related Phenomena, MIAS, (1997), p. 87.
- A6.3 Baxter, G J, Furu, T, Whiteman, J A and Sellars, C M: Mater. Sci. Forum, 217-222 (1996), 459.
- A6.4 Urcola, J J and Sellars, C M: Acta Metall., 35 (1987), 2649.
- A6.5 Immarigeon, J-P and Jonas, J J: Acta Metall., 19 (1971), 1053.
- A6.6 Karjalainen, L P and Somani, M C: Effect of strain reversals on recrystallization kinetics and grain size in hot-deformed austenite, Proc. Intern. Conf. Recrystallization and Grain Growth, eds, G. Gottstein and D.A. Molodov, August 27-31, 2001, Aachen, Germany, vol. 2, pp779-784.
- A6.7 Karjalainen, L P, Porter, D and Peura, P: Recrystallization of Ti-microalloyed steels under constant and varying deformation conditions. The Fourth International Conference on Recrystallization and Related Phenomena, July 13-16, 1999. Eds. T. Sakai and H.G. Suzuki. Proceedings of ReX'99 (JIMIS-10), 1999. pp721 - 726.
- A6.8 Karjalainen, L P, Somani, M C, Peura, P A and Porter, D A: Effects of strain rate changes and strain path on flow stress and recrystallisation kinetics in Nb-bearing microalloy steels, Thermomechanical Processing of Steels, 24-26 May 2000, London, UK, pp130-139.
- A6.9 Bartolomé, R, Astiazarán, J I, Iza-Mendia, A and Gutiérrez, I: Mechanical and microstructural effect of the strain reversal on a microalloyed steel, Thermomechanical Processing of Steels, 24-26 May 2000, London, UK, pp221-230.
- A6.10 Jorge-Badiola, D, Bartolomé, R, Martin, S and Gutiérrez, I: Effect of the strain reversal on both the recrystallization and the strain-induced precipitation in a Nb-microalloyed steel, Intern. Conf. Thermomechanical Processing: Mechanisms, Microstructure and Control, 23-26 June, 2002, Sheffield, UK.
- A6.11 Bartolomé, R, Jorge-Badiola, D, Astiazarán, J I and Gutiérrez, I: Mater. Sci. Eng., A344, 2003, pp323-330.
- A6.12 Perttula, J S and Karjalainen, L P: Mat. Sci. Techn., 1998, 14, (7), pp626-630.
- A6.13 Jansen, E F M, Geisler, S, De Paepe, A, Underhill, R and Anelli, E: New hot rolled deep drawing steel qualities, Final Report, 2001, EUR 19889 EN.
- A6.14 Sellars, C M and Zhu, Q: Mater. Sci. Eng. A280, 2000, pp1-7.
- A6.15 Furu, T, Shercliff, H F, Sellars, C M and Ashby, M F: Mater. Sci. Forum, 1996, 217-222, pp453-458.

- A6.16 Karjalainen, L P and Perttula, J: Characteristics of static and metadynamic recrystallization and strain accumulation in hot-deformed austenite as revealed by the stress relaxation method, *ISIJ International*, 1996, 36, (6), pp729-736.
- A6.17 Barnett, M R, Kelly, G L and Hodgson, P D: *Metall. Mater. Trans. A*, 2002, 33A, (7), pp1893-1900.
- A6.18 Zhu, Q and Sellars, C M: Effect of deformation paths on static recrystallisation behaviour of an Al-2Mg alloy, *Proc. ReX'96, Third Intern. Conf. recrystallization and related phenomena*, ed. McNelley, T R, Monterey, Ca, USA, 1996, pp195-202.
- A6.19 Bergström, Y: *The Plastic Deformation of Metals – A Dislocation Model and its Applicability*, Royal Institute of Technology, Tech. Report, Stockholm, 1982.
- A6.20 Bianchi, J H and Karjalainen, L P: Modelling of dynamic and metadynamic recrystallisation during bar rolling of a medium carbon spring steel, *J. Materials Processing Technology*, Elsevier, (2005), vol. 160, No. 3, pp267-277.

Table A6.1: Chemical composition of the steels tested (wt. %)

Steel	C	Si	Mn	Al	Ni	V	Ti	Nb	B(ppm)
Nb	0.15	0.3	1.42	0.037	0.03	0.011		0.033	
hTi-Nb	0.07	0.20	1.80	0.028	0.03	0.007	0.130	0.067	3
42MnV7	0.40	0.22	1.68	0.036	0.04	0.083	0.018	0.003	4
Low-C	0.092	0.190	0.450	0.026					
Ti-IF	0.002	0.004	0.128	0.024	0.022		0.071		
Nb-Ti-IF	0.003	0.010	0.164	0.036	0.032	0.003	0.020	0.037	
Armco	0.014		0.020						

Table A6.2: Results of temperature history tests

Steel: hTi-Nb (0.071C-1.8Mn-0.067Nb-0.13Ti-0.007V)

RH 1250°C/ 5 min, followed by cooling at 2°C/s to T_{def} ,

15 s hold, followed by compression.

Grain size: $\approx 29 \mu\text{m}$, total strain: ≈ 0.2

No.	Test details	Strain rate (s^{-1})	Relax. stress (MPa)	n	t5 (s)	t50 (s)	t95 (s)
1	Isothermal 1025°C/0.2	0.1	80	1.4	5.2	34	96
2	Isothermal 1050°C/0.2	0.1	80	1.4	5.3	36	100
			83	1.4	6.0	38	110
			77	1.7	6.3	30	71
3	Isothermal 1075°C/0.2	0.1	71	1.6	4.0	21	52
		0.1	70	1.6	3.5	18	46
4	Isothermal 1100°C/0.2	0.1	62	1.6	2.3	12	30
5	Isothermal 1125°C/0.2	0.1	69	1.4	1.0	6.7	20
			63.5	1.9	1.7	6.2	13
6	Isothermal 1150°C/0.2	0.1	59.5	1.5	1.4	7.8	20
			57	1.9	1.2	4.8	10
7	Isothermal 1100°C/0.2	0.01	49	1.7	4.3	20	48
8	Deformation during heating 1000→1100°C/0.2	0.1	67	1.7	3.1	14.4	34

9	Deformation during heating 1000→1100°C/0.19, isothermal 1100°C/0.01	0.1 0.1	65.5 67	1.5 1.7	1.8 3	10.2 14	28 32
10	Deformation during heating 1000→1100°C/0.18, isothermal 1100°C/0.02	0.1 0.1	65 67	1.6 1.6	2.9 2.7	15 13.5	38 33
11	Deformation during heating 1050→1150°C/0.19, isothermal 1150°C/0.01	0.1	55.5	1.4	1.2	7.7	21.5
12	Isothermal at 900°C/0.16, 900→1100°C/0.04*	0.01	74				
13	Isothermal at 900°C/0.15, 900→1100°C/0.04*, isothermal 1100°C/0.01	0.01 0.01	72 69				
14	Isothermal at 1000°C/0.16, 1000→1100°C/0.02*, isothermal 1100°C/0.02	0.01 0.01	54 56.5	1.6 1.6	3.9 4.6	19.5 23	50 59

→ Concurrent heating at 50-55.5°C/s during deformation

* Heating started while the deformation is being continued

Steel: 42MnV7 (0.398C-1.68Mn-0.018Ti-0.083V)

RH 1250°C/5 min, followed by cooling at 2°C/s to T_{def} ,

15 s hold, followed by compression.

Grain size: $\approx 125 \mu\text{m}$, strain: ≈ 0.2

No.	Test details	Strain rate (s ⁻¹)	Relax stress (MPa)	n	t5 (s)	t50 (s)	t95 (s)
1	Isothermal 1100°C/0.2	0.1	58	1.6	0.68	3.3	8.2
2	Isothermal 1100°C/0.2	0.1	59	2.0	0.50	1.75	3.8
3	Isothermal 1100°C/0.2	0.01					
4	Deformation during heating 1000→1100°C/0.2	0.1	61	1.6	0.38	2	5.0
5	Isothermally at 900°C/0.16, 900→1100°C/0.04*	0.01	46	1.1	0.29	3	11.5

*concurrent heating at 50°C/s during deformation

Table A6.3: Cont...

Sl. No.	Name of the Candidate	Registration Number	Grade	Score	Percentage	Rank	Remarks
1	ABHIRAM K	1001	10	95	95.00	1	
2	ADARSH K	1002	10	90	90.00	2	
3	ADITHYAN K	1003	10	85	85.00	3	
4	ADITHYAN K	1004	10	80	80.00	4	
5	ADITHYAN K	1005	10	75	75.00	5	
6	ADITHYAN K	1006	10	70	70.00	6	
7	ADITHYAN K	1007	10	65	65.00	7	
8	ADITHYAN K	1008	10	60	60.00	8	
9	ADITHYAN K	1009	10	55	55.00	9	
10	ADITHYAN K	1010	10	50	50.00	10	
11	ADITHYAN K	1011	10	45	45.00	11	
12	ADITHYAN K	1012	10	40	40.00	12	
13	ADITHYAN K	1013	10	35	35.00	13	
14	ADITHYAN K	1014	10	30	30.00	14	
15	ADITHYAN K	1015	10	25	25.00	15	
16	ADITHYAN K	1016	10	20	20.00	16	
17	ADITHYAN K	1017	10	15	15.00	17	
18	ADITHYAN K	1018	10	10	10.00	18	
19	ADITHYAN K	1019	10	5	5.00	19	
20	ADITHYAN K	1020	10	0	0.00	20	

Table A6.4: Cont...

[illegible]

Sl. No.	Name of the Candidate	Roll No.	Grade	Score	Remarks
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Table A6.5: Strain rate change tests performed for hti-nb steel

A6.5.1 Strain rate

T17 test

The T17 test is a standard test for hti-nb steel, performed at a constant strain rate of 0.001 s⁻¹ and a constant temperature of 17°C.

Test No.	Strain rate	Strain rate	Strain rate	Strain rate	Strain rate
T17-1	0.001	0.001	0.001	0.001	0.001
T17-2	0.001	0.001	0.001	0.001	0.001
T17-3	0.001	0.001	0.001	0.001	0.001
T17-4	0.001	0.001	0.001	0.001	0.001
T17-5	0.001	0.001	0.001	0.001	0.001
T17-6	0.001	0.001	0.001	0.001	0.001
T17-7	0.001	0.001	0.001	0.001	0.001
T17-8	0.001	0.001	0.001	0.001	0.001

A6.5.2 Strain rate

The A6.5.2 test is a standard test for hti-nb steel, performed at a constant strain rate of 0.001 s⁻¹ and a constant temperature of 17°C.

Test No.	Strain rate	Strain rate	Strain rate	Strain rate	Strain rate
A6.5.2-1	0.001	0.001	0.001	0.001	0.001
A6.5.2-2	0.001	0.001	0.001	0.001	0.001

A6.5.3 Strain rate

The A6.5.3 test is a standard test for hti-nb steel, performed at a constant strain rate of 0.001 s⁻¹ and a constant temperature of 17°C. The test is performed at a constant strain rate of 0.001 s⁻¹ and a constant temperature of 17°C.

Test No.	Strain rate	Strain rate	Strain rate	Strain rate	Strain rate	Strain rate	Strain rate	Strain rate
A6.5.3-1	0.001	0.001	0.001	0.001	0.001	0.001	0.001	0.001
A6.5.3-2	0.001	0.001	0.001	0.001	0.001	0.001	0.001	0.001
A6.5.3-3	0.001	0.001	0.001	0.001	0.001	0.001	0.001	0.001
A6.5.3-4	0.001	0.001	0.001	0.001	0.001	0.001	0.001	0.001
A6.5.3-5	0.001	0.001	0.001	0.001	0.001	0.001	0.001	0.001
A6.5.3-6	0.001	0.001	0.001	0.001	0.001	0.001	0.001	0.001
A6.5.3-7	0.001	0.001	0.001	0.001	0.001	0.001	0.001	0.001
A6.5.3-8	0.001	0.001	0.001	0.001	0.001	0.001	0.001	0.001
A6.5.3-9	0.001	0.001	0.001	0.001	0.001	0.001	0.001	0.001

A6.5.4 Strain rate

The A6.5.4 test is a standard test for hti-nb steel, performed at a constant strain rate of 0.001 s⁻¹ and a constant temperature of 17°C.

Test No.	Strain rate	Strain rate	Strain rate	Strain rate	Strain rate	Strain rate	Strain rate	Strain rate
A6.5.4-1	0.001	0.001	0.001	0.001	0.001	0.001	0.001	0.001
A6.5.4-2	0.001	0.001	0.001	0.001	0.001	0.001	0.001	0.001
A6.5.4-3	0.001	0.001	0.001	0.001	0.001	0.001	0.001	0.001
A6.5.4-4	0.001	0.001	0.001	0.001	0.001	0.001	0.001	0.001

The A6.5.4 test is a standard test for hti-nb steel, performed at a constant strain rate of 0.001 s⁻¹ and a constant temperature of 17°C.

A6.5.5 Strain rate

The A6.5.5 test is a standard test for hti-nb steel, performed at a constant strain rate of 0.001 s⁻¹ and a constant temperature of 17°C.

Test No.	Strain rate	Strain rate	Strain rate	Strain rate	Strain rate	Strain rate	Strain rate	Strain rate
A6.5.5-1	0.001	0.001	0.001	0.001	0.001	0.001	0.001	0.001
A6.5.5-2	0.001	0.001	0.001	0.001	0.001	0.001	0.001	0.001
A6.5.5-3	0.001	0.001	0.001	0.001	0.001	0.001	0.001	0.001

Table A6.6: Strain rate change tests performed for low-c steel

Lower strain rate

For 1000/1000 only, loaded at 5000 to 10000, then held
loading to 10000 at 10000, then held, followed by elastic unloading

Test = 70000

for strain rate

10000 to 10000

Test No.	Material	Temperature (°C)	Strain rate (s ⁻¹)	Stress (MPa)	Strain (mm/mm)	Time between stresses, s	Residual stress N	Residual strain mm
1	Steel 1.50	700	1	0.0	0.0	0.0	0.00	0.00
2	Steel 1.50	700	1	0.0	0.0	1	0.00	0.00
3	Steel 1.50	700	1	0.0	0.0	2	0.00	0.00
4	Steel 1.50	700	1	0.0	0.0	3	0.00	0.00
5	Steel 1.50	700	1	0.0	0.0	4	0.00	0.00
6	Steel 1.50	700	1	0.0	0.0	5	0.00	0.00
7	Steel 1.50	700	1	0.0	0.0	6	0.00	0.00
8	Steel 1.50	700	1	0.0	0.0	7	0.00	0.00
9	Steel 1.50	700	1	0.0	0.0	8	0.00	0.00
10	Steel 1.50	700	1	0.0	0.0	9	0.00	0.00
11	Steel 1.50	700	1	0.0	0.0	10	0.00	0.00
12	Steel 1.50	700	1	0.0	0.0	11	0.00	0.00
13	Steel 1.50	700	1	0.0	0.0	12	0.00	0.00
14	Steel 1.50	700	1	0.0	0.0	13	0.00	0.00
15	Steel 1.50	700	1	0.0	0.0	14	0.00	0.00
16	Steel 1.50	700	1	0.0	0.0	15	0.00	0.00
17	Steel 1.50	700	1	0.0	0.0	16	0.00	0.00
18	Steel 1.50	700	1	0.0	0.0	17	0.00	0.00
19	Steel 1.50	700	1	0.0	0.0	18	0.00	0.00
20	Steel 1.50	700	1	0.0	0.0	19	0.00	0.00
21	Steel 1.50	700	1	0.0	0.0	20	0.00	0.00
22	Steel 1.50	700	1	0.0	0.0	21	0.00	0.00
23	Steel 1.50	700	1	0.0	0.0	22	0.00	0.00
24	Steel 1.50	700	1	0.0	0.0	23	0.00	0.00
25	Steel 1.50	700	1	0.0	0.0	24	0.00	0.00
26	Steel 1.50	700	1	0.0	0.0	25	0.00	0.00
27	Steel 1.50	700	1	0.0	0.0	26	0.00	0.00
28	Steel 1.50	700	1	0.0	0.0	27	0.00	0.00
29	Steel 1.50	700	1	0.0	0.0	28	0.00	0.00
30	Steel 1.50	700	1	0.0	0.0	29	0.00	0.00
31	Steel 1.50	700	1	0.0	0.0	30	0.00	0.00
32	Steel 1.50	700	1	0.0	0.0	31	0.00	0.00
33	Steel 1.50	700	1	0.0	0.0	32	0.00	0.00
34	Steel 1.50	700	1	0.0	0.0	33	0.00	0.00
35	Steel 1.50	700	1	0.0	0.0	34	0.00	0.00
36	Steel 1.50	700	1	0.0	0.0	35	0.00	0.00
37	Steel 1.50	700	1	0.0	0.0	36	0.00	0.00
38	Steel 1.50	700	1	0.0	0.0	37	0.00	0.00
39	Steel 1.50	700	1	0.0	0.0	38	0.00	0.00
40	Steel 1.50	700	1	0.0	0.0	39	0.00	0.00
41	Steel 1.50	700	1	0.0	0.0	40	0.00	0.00
42	Steel 1.50	700	1	0.0	0.0	41	0.00	0.00
43	Steel 1.50	700	1	0.0	0.0	42	0.00	0.00
44	Steel 1.50	700	1	0.0	0.0	43	0.00	0.00
45	Steel 1.50	700	1	0.0	0.0	44	0.00	0.00
46	Steel 1.50	700	1	0.0	0.0	45	0.00	0.00
47	Steel 1.50	700	1	0.0	0.0	46	0.00	0.00
48	Steel 1.50	700	1	0.0	0.0	47	0.00	0.00
49	Steel 1.50	700	1	0.0	0.0	48	0.00	0.00
50	Steel 1.50	700	1	0.0	0.0	49	0.00	0.00
51	Steel 1.50	700	1	0.0	0.0	50	0.00	0.00
52	Steel 1.50	700	1	0.0	0.0	51	0.00	0.00
53	Steel 1.50	700	1	0.0	0.0	52	0.00	0.00
54	Steel 1.50	700	1	0.0	0.0	53	0.00	0.00
55	Steel 1.50	700	1	0.0	0.0	54	0.00	0.00
56	Steel 1.50	700	1	0.0	0.0	55	0.00	0.00
57	Steel 1.50	700	1	0.0	0.0	56	0.00	0.00
58	Steel 1.50	700	1	0.0	0.0	57	0.00	0.00
59	Steel 1.50	700	1	0.0	0.0	58	0.00	0.00
60	Steel 1.50	700	1	0.0	0.0	59	0.00	0.00
61	Steel 1.50	700	1	0.0	0.0	60	0.00	0.00
62	Steel 1.50	700	1	0.0	0.0	61	0.00	0.00
63	Steel 1.50	700	1	0.0	0.0	62	0.00	0.00
64	Steel 1.50	700	1	0.0	0.0	63	0.00	0.00
65	Steel 1.50	700	1	0.0	0.0	64	0.00	0.00
66	Steel 1.50	700	1	0.0	0.0	65	0.00	0.00
67	Steel 1.50	700	1	0.0	0.0	66	0.00	0.00
68	Steel 1.50	700	1	0.0	0.0	67	0.00	0.00
69	Steel 1.50	700	1	0.0	0.0	68	0.00	0.00
70	Steel 1.50	700	1	0.0	0.0	69	0.00	0.00
71	Steel 1.50	700	1	0.0	0.0	70	0.00	0.00
72	Steel 1.50	700	1	0.0	0.0	71	0.00	0.00
73	Steel 1.50	700	1	0.0	0.0	72	0.00	0.00
74	Steel 1.50	700	1	0.0	0.0	73	0.00	0.00
75	Steel 1.50	700	1	0.0	0.0	74	0.00	0.00
76	Steel 1.50	700	1	0.0	0.0	75	0.00	0.00
77	Steel 1.50	700	1	0.0	0.0	76	0.00	0.00
78	Steel 1.50	700	1	0.0	0.0	77	0.00	0.00
79	Steel 1.50	700	1	0.0	0.0	78	0.00	0.00
80	Steel 1.50	700	1	0.0	0.0	79	0.00	0.00
81	Steel 1.50	700	1	0.0	0.0	80	0.00	0.00
82	Steel 1.50	700	1	0.0	0.0	81	0.00	0.00
83	Steel 1.50	700	1	0.0	0.0	82	0.00	0.00
84	Steel 1.50	700	1	0.0	0.0	83	0.00	0.00
85	Steel 1.50	700	1	0.0	0.0	84	0.00	0.00
86	Steel 1.50	700	1	0.0	0.0	85	0.00	0.00
87	Steel 1.50	700	1	0.0	0.0	86	0.00	0.00
88	Steel 1.50	700	1	0.0	0.0	87	0.00	0.00
89	Steel 1.50	700	1	0.0	0.0	88	0.00	0.00
90	Steel 1.50	700	1	0.0	0.0	89	0.00	0.00
91	Steel 1.50	700	1	0.0	0.0	90	0.00	0.00
92	Steel 1.50	700	1	0.0	0.0	91	0.00	0.00
93	Steel 1.50	700	1	0.0	0.0	92	0.00	0.00
94	Steel 1.50	700	1	0.0	0.0	93	0.00	0.00
95	Steel 1.50	700	1	0.0	0.0	94	0.00	0.00
96	Steel 1.50	700	1	0.0	0.0	95	0.00	0.00
97	Steel 1.50	700	1	0.0	0.0	96	0.00	0.00
98	Steel 1.50	700	1	0.0	0.0	97	0.00	0.00
99	Steel 1.50	700	1	0.0	0.0	98	0.00	0.00
100	Steel 1.50	700	1	0.0	0.0	99	0.00	0.00

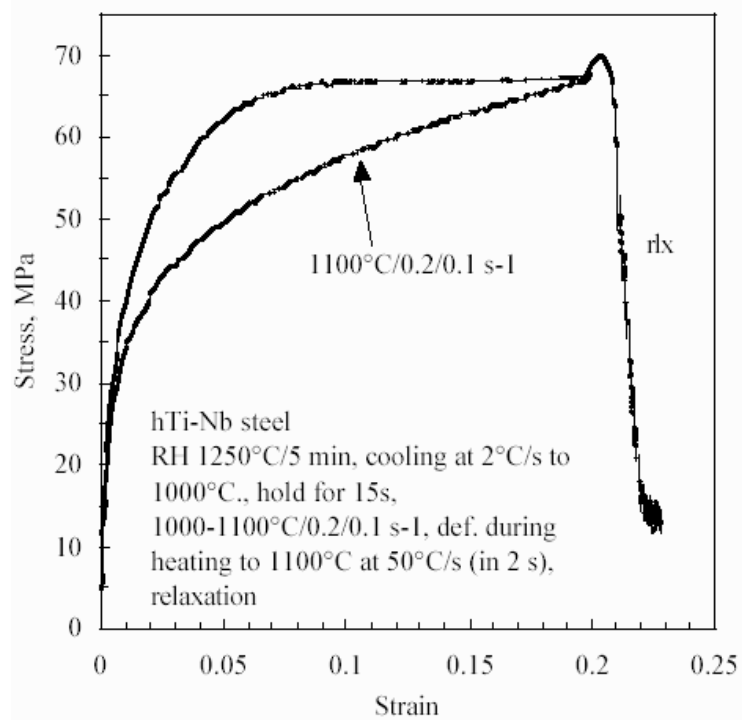
Table A6.7: Data concerning the deformation cells in the if-steel and armco iron

Material	Test type	No of specimens (TEM/BF)	No. of subgrains	Average subgrain size (μm)
Ti-IF	Const. ϵ'	3	~ 100	1.4
Ti-IF	SRC	3	~ 425	1.8
Armco	C	5	~ 150	2.8
Armco	T/C	5	~ 110	4.1

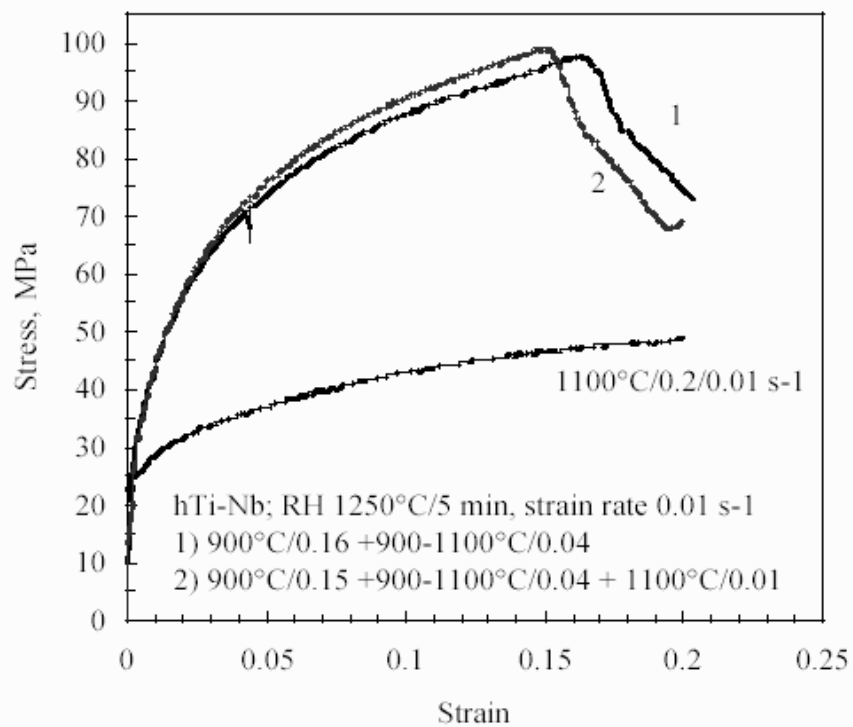
Material	Test type	No. of subgrains	Average subgrain size (μm)	SEM/EBSD
Ti-IF	Const. ϵ'	~ 2000	2.2	yes
Ti-IF	SRC	~ 2400	2.8	yes
Armco	C	~ 2700	4.6	yes
Armco	T/C	~ 2200	4.5	yes

Table A6.8: Experiments for modelling the transient behavior of ferrite.
Some predicted data are also given.

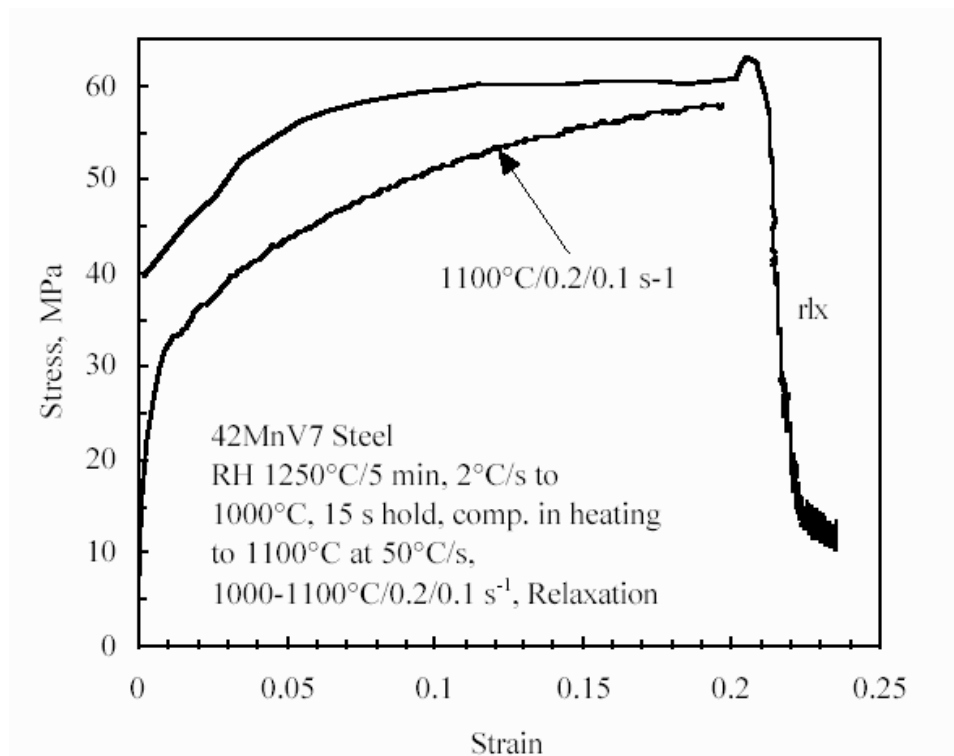
Material: Low C steel (0.092C-0.19Si-0.45Mn-0.008V)											
Phase structure: BCC ferrite											
Proc/ RH 1000°C/5 min, at 5°C/s to 650°C, 10 s hold, to 700°C at 10°C/s, 10 s hold, followed by compr											
Temperature: 700°C											
$\alpha = 0.88$, $G = 58$ GPa, $b = 25$ nm											
Test	Schedule	σ ($\epsilon=0.6$)	σ_0	σ_0 SRC	U ($=d\rho/d\epsilon$)	Ω	Ω	Est.	ρ_0	ρ_i ($\epsilon=0.6$)	ρ_m ($\epsilon=0.6$)
	str rate/str	(MNm ⁻²)	(MNm ⁻²)		(m ⁻²)		at SRC	trans ϵ	(m ⁻²)	(m ⁻²)	*x (m ⁻²)
1	1	216	35		2.3E+15	11.5			2E+12	2.01E+14	1
2	0.1	173	29		1.72E+15	13			2E+12	1.27E+14	0.12
3	0.01	125.5	25		1.1E+15	18			2E+12	6.20E+13	0.014
4	0.001	90	20		8E+14	25			2E+12	3.01E+13	0.00175
5	1/0.6 + 0.1/0.4	216	35		2.3E+15	11.5			2E+12	2.01E+14	1
			13.8 to 29	13.8	1.72E+15	19 to 13	19.3	0.122	2.01E+14		0.254
6	1/0.6+0.01/0.4	216	35		2.3E+15	11.5			2E+12	2.01E+14	1
			3.7 to 25	3.7	1.1E+15	42 to 18	42.2	0.18	2.01E+14		0.0949
7	1/0.6+0.001/0.4	216	35		2.3E+15	11.5			2E+12	2.01E+14	1
			1.1 to 20	1.1	8E+14	86 to 25	86.3	0.27	2.01E+14		0.0336



(a)

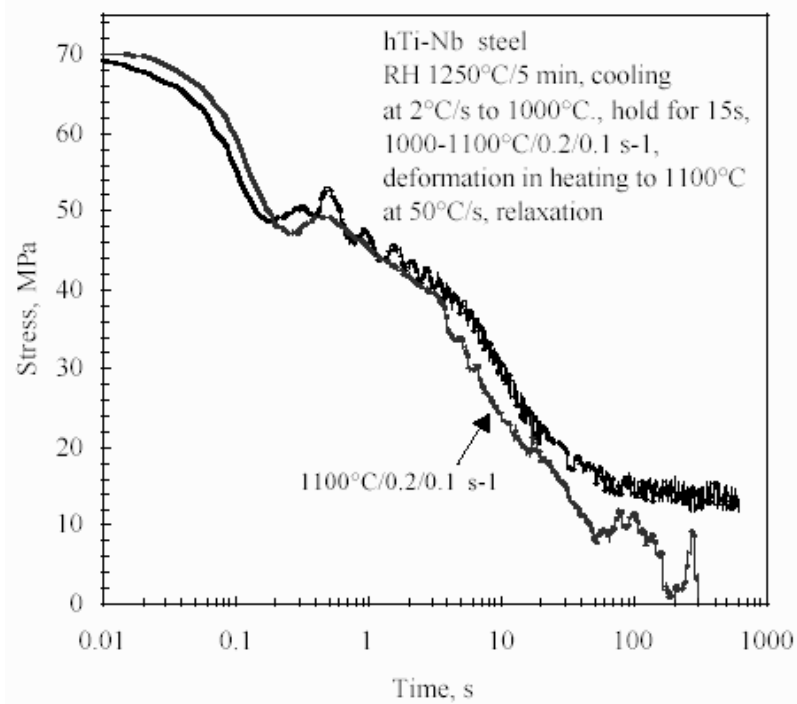


(b)

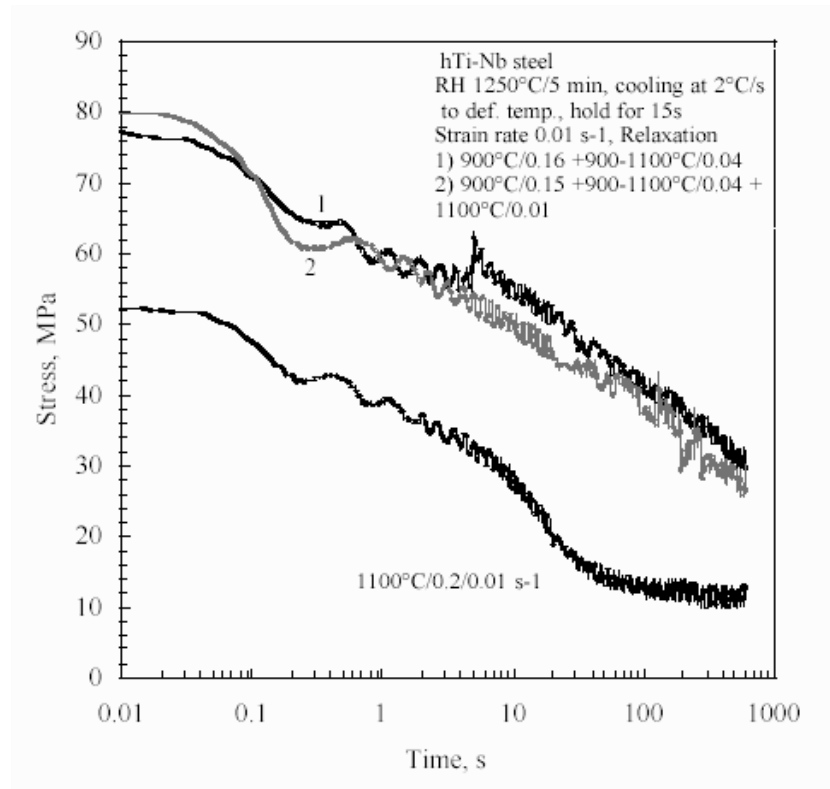


(c)

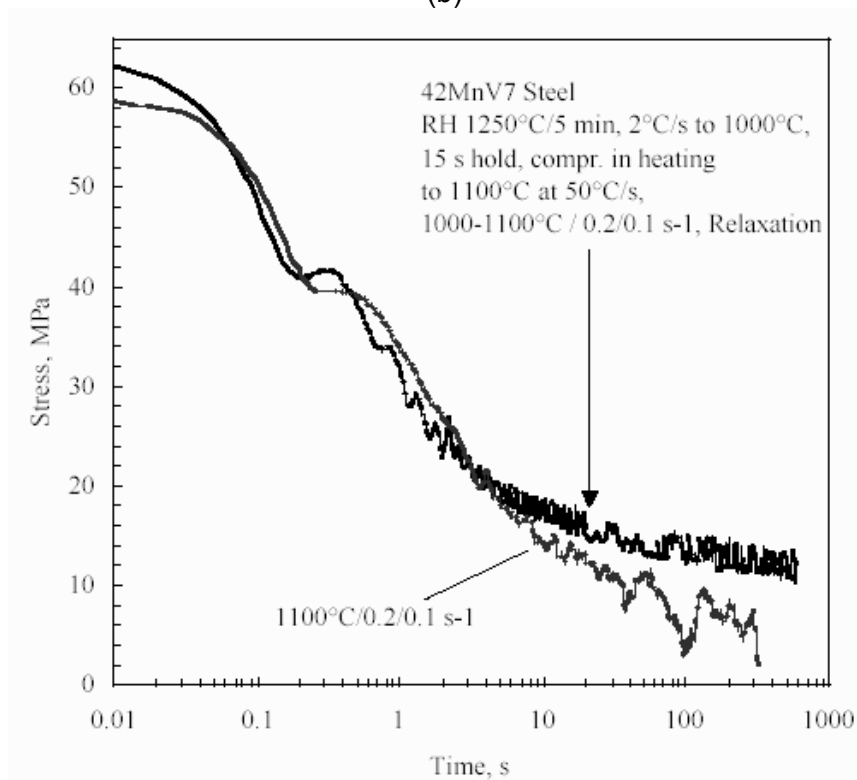
Fig. A6.1(a-c): Typical flow stress curves in isothermal and temperature change tests



(a)

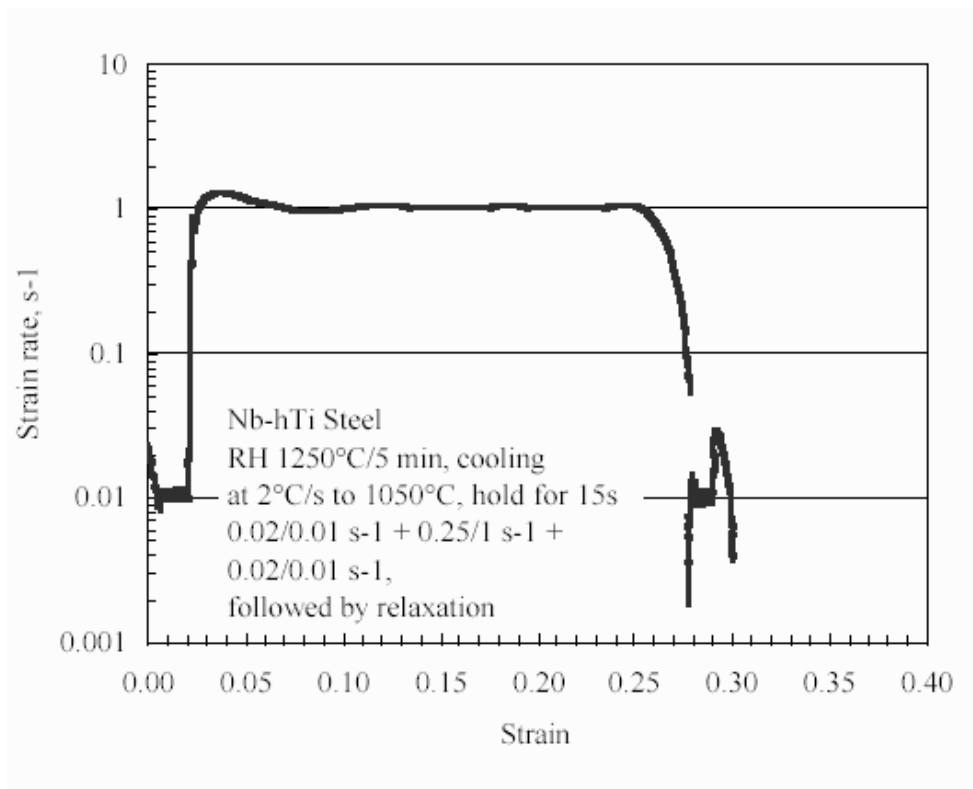


(b)

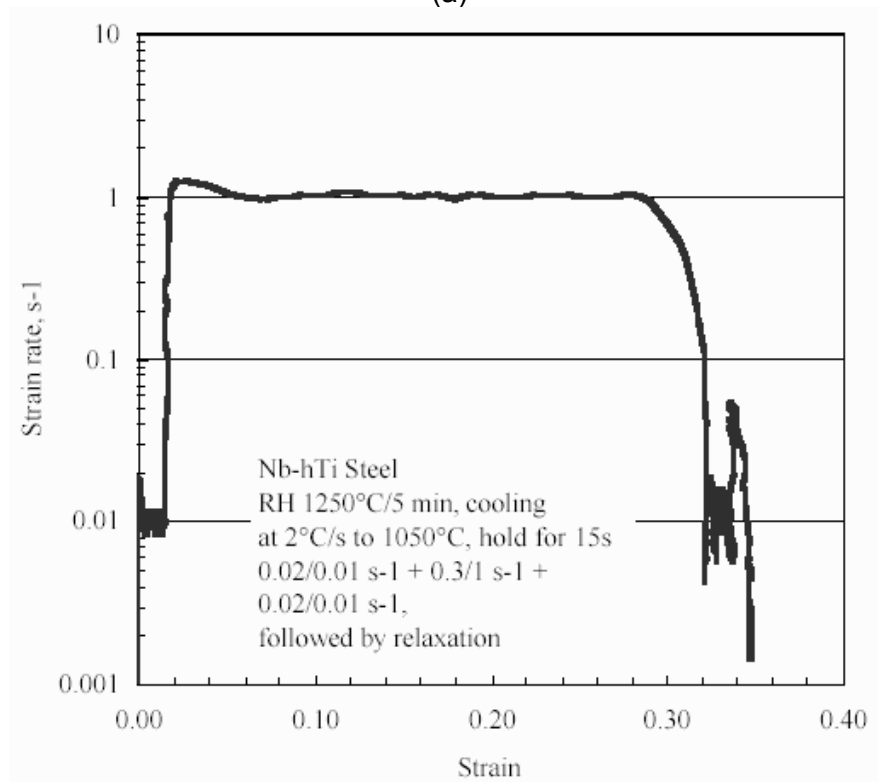


(c)

Fig. A6.2(a-c): Typical relaxation curves in isothermal and temperature change tests

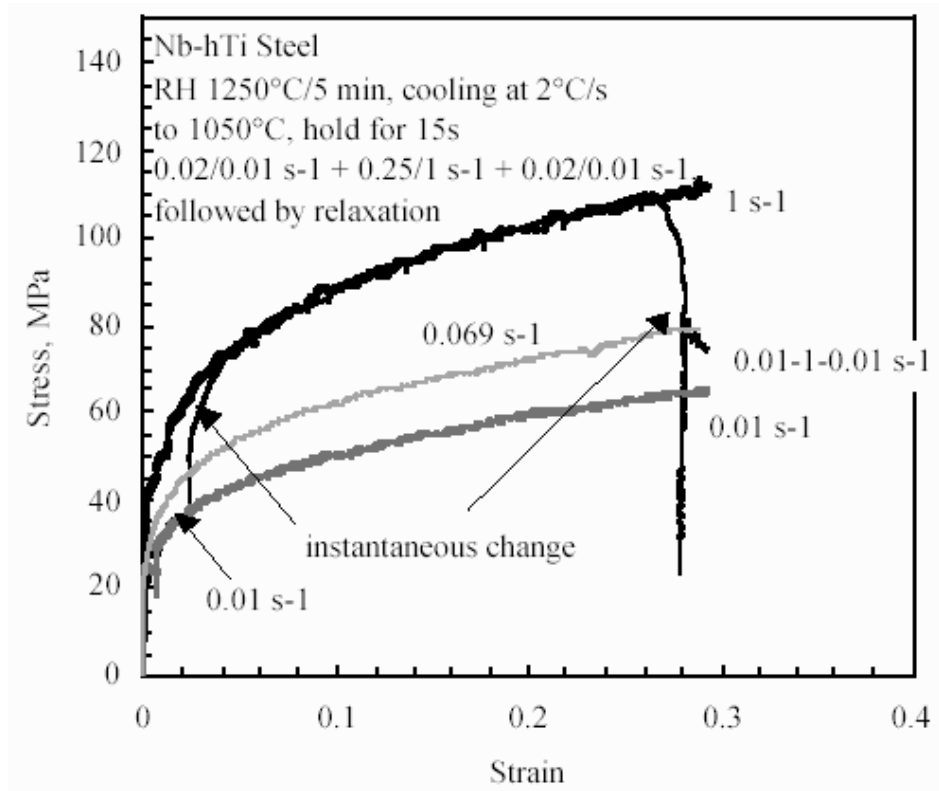


(a)

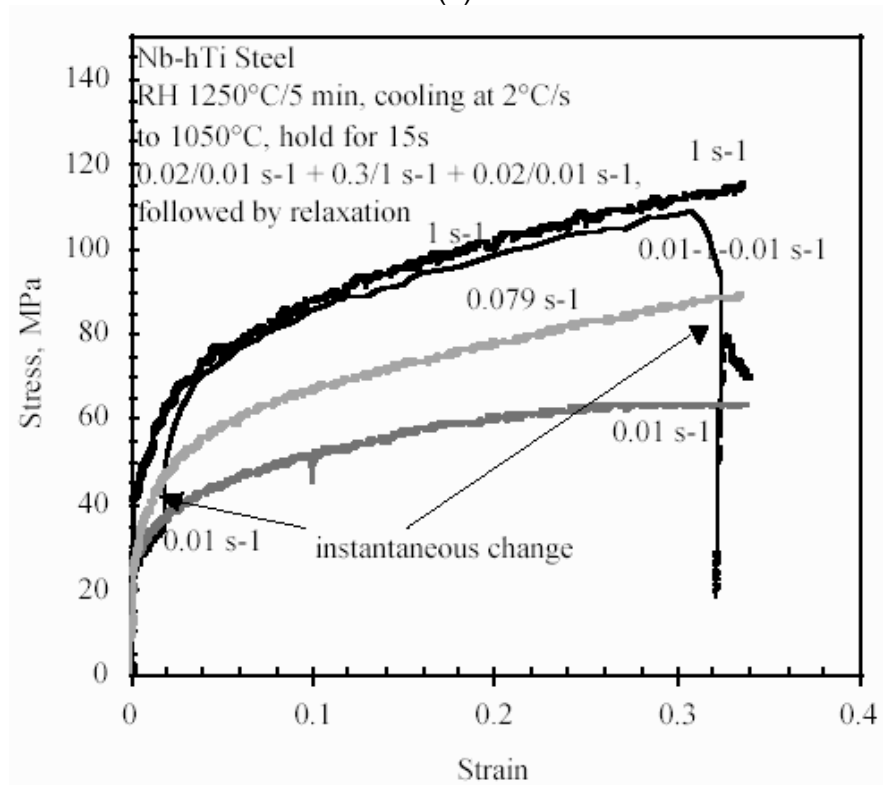


(b)

Fig. A6.3(a and b): Strain rate history in tests



(a)



(b)

Fig. A6.4(a and b): Influence of strain rate history on flow stress.

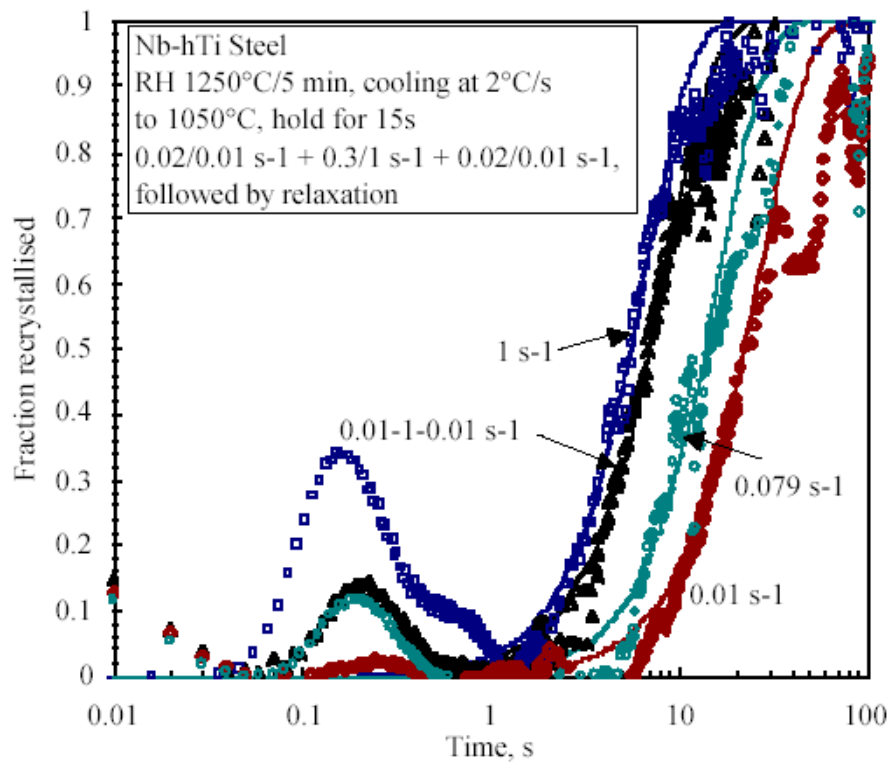
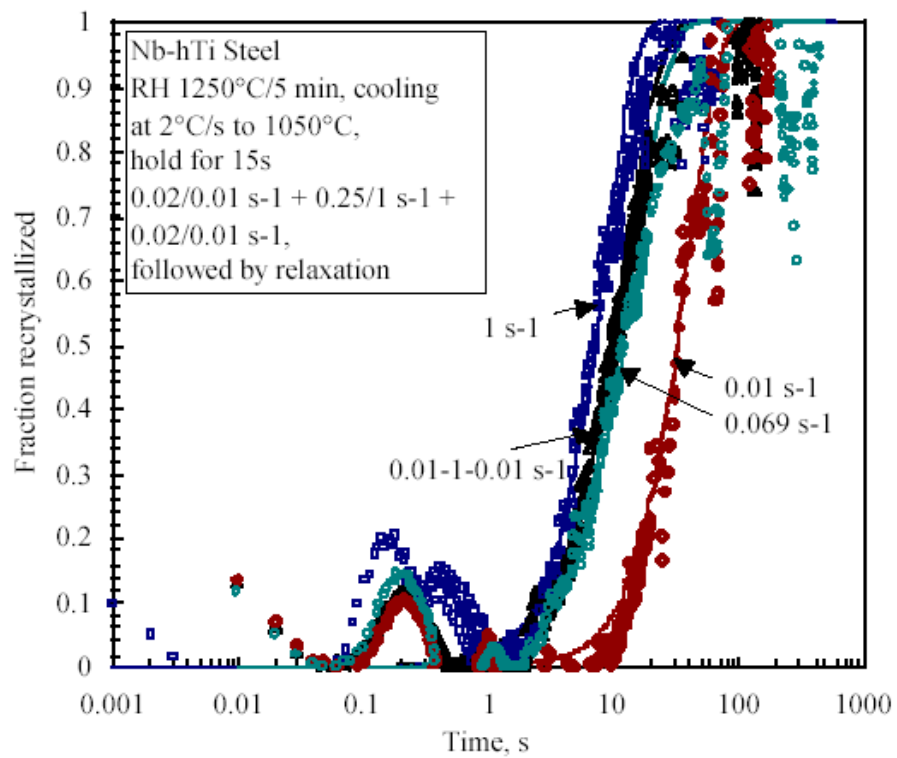


Fig. A6.5: Effect of strain rate history on fractional softening curves.

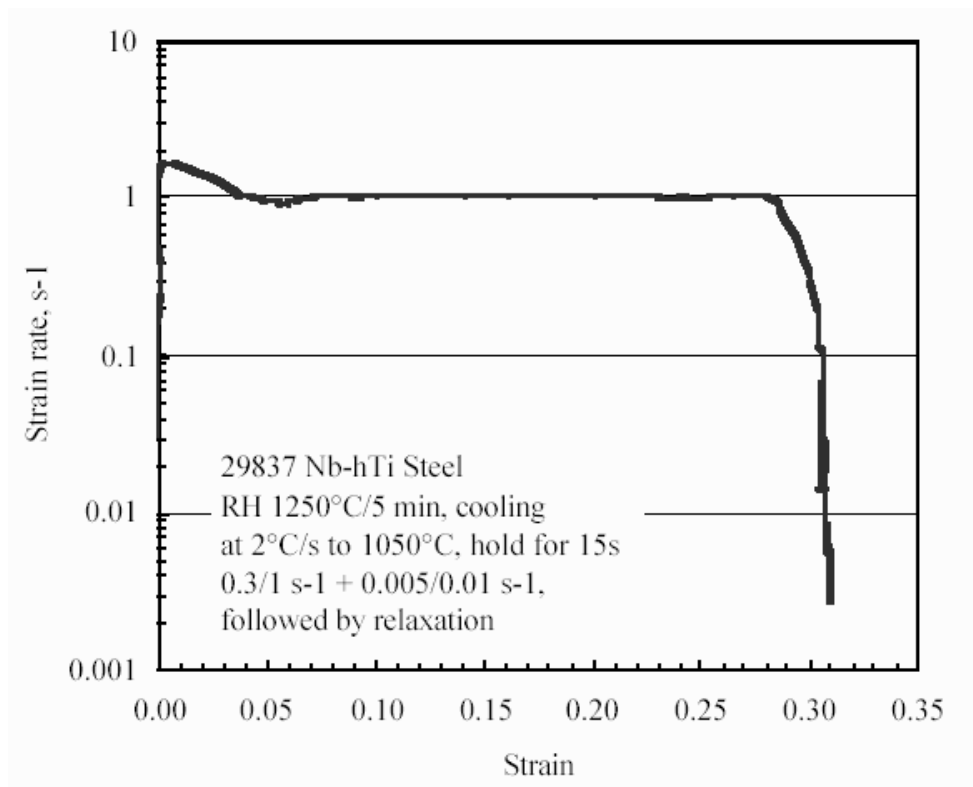
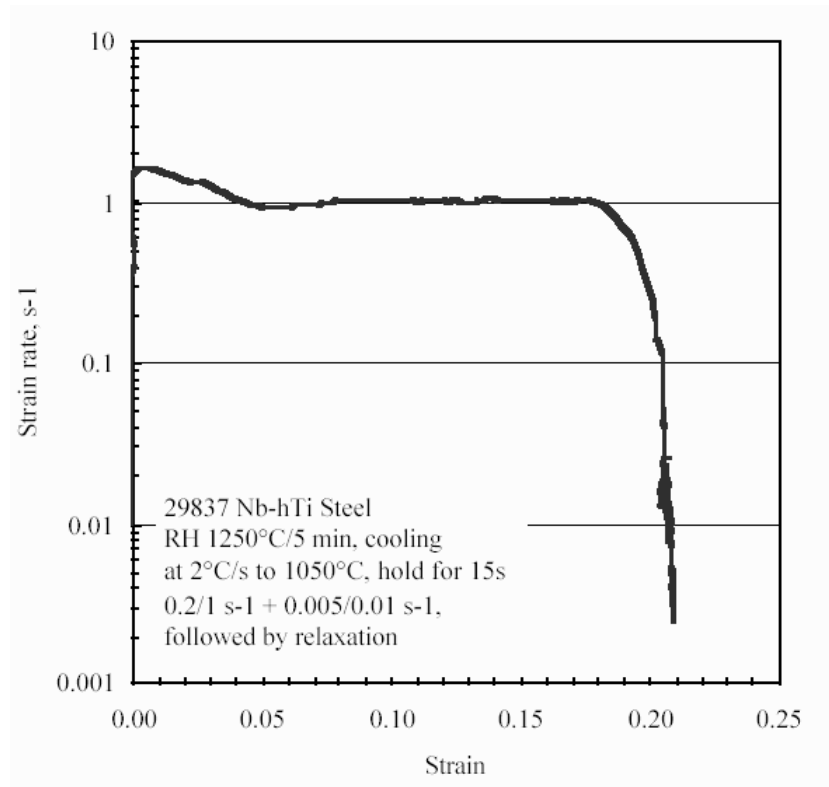


Fig. A6.6: Strain rate history in the second group tests.

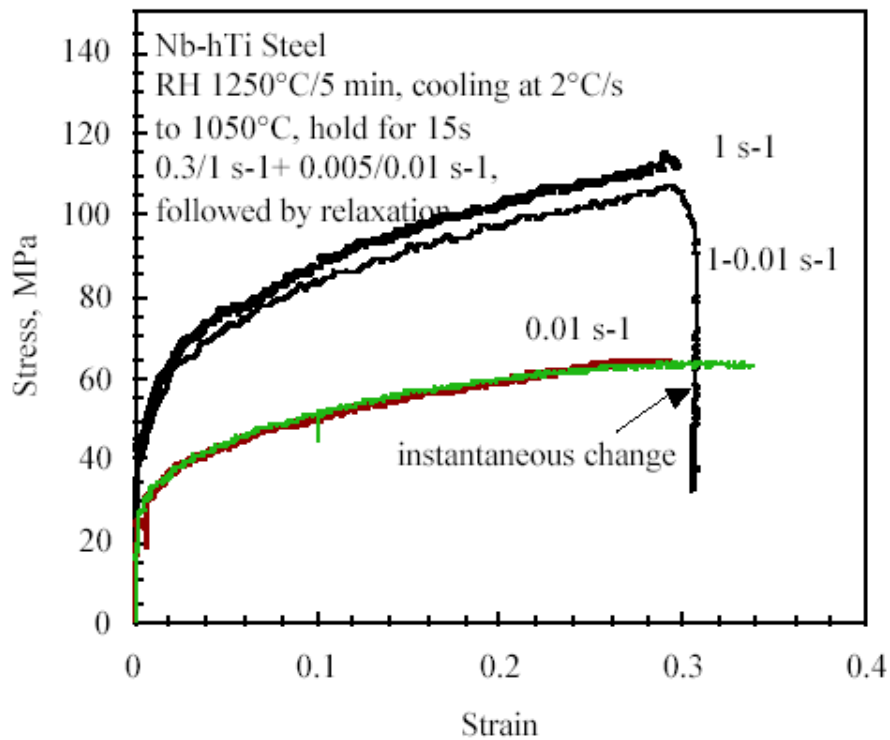
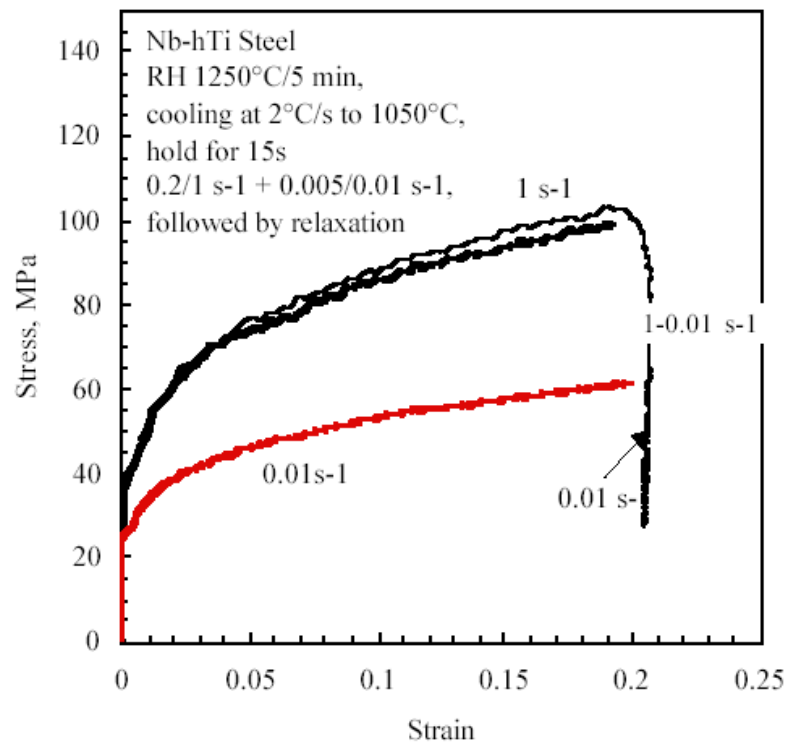


Fig. A6.7: Influence of strain rate history on flow stress.

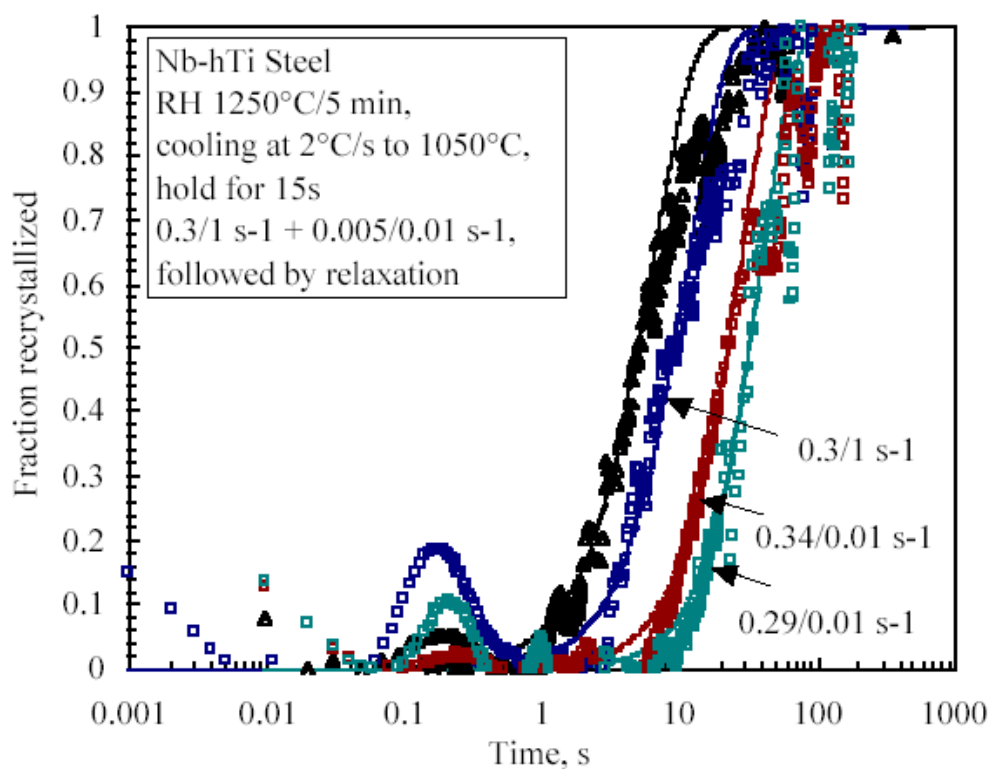
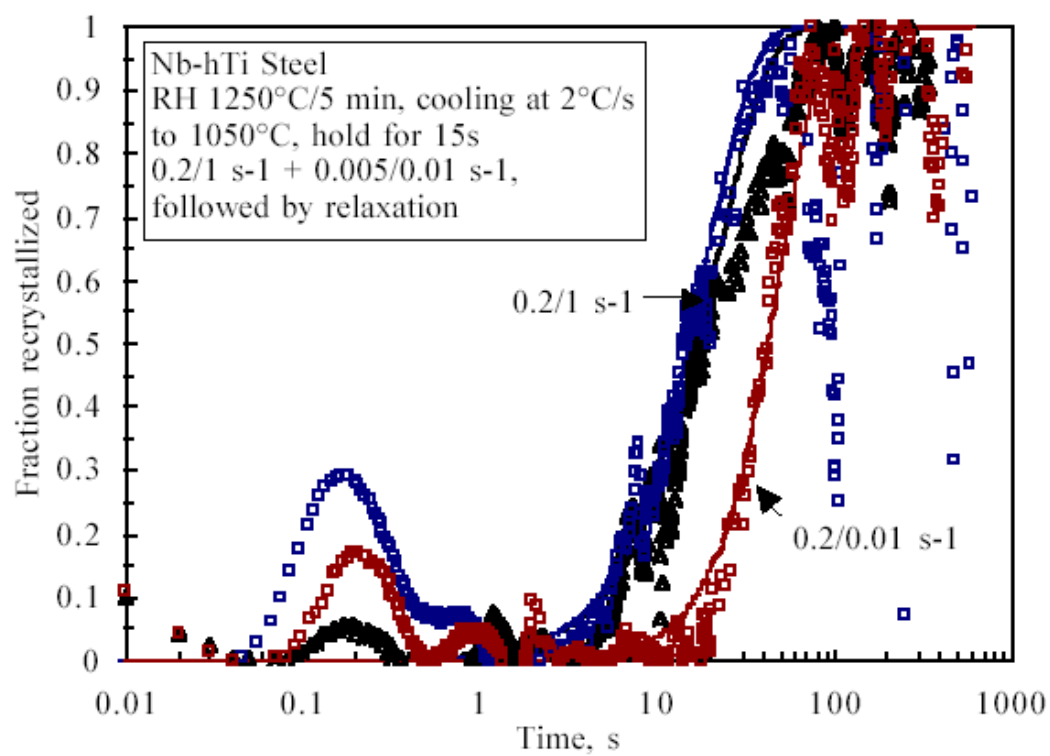


Fig. A6.8: Effect of strain rate history on fractional softening curves.

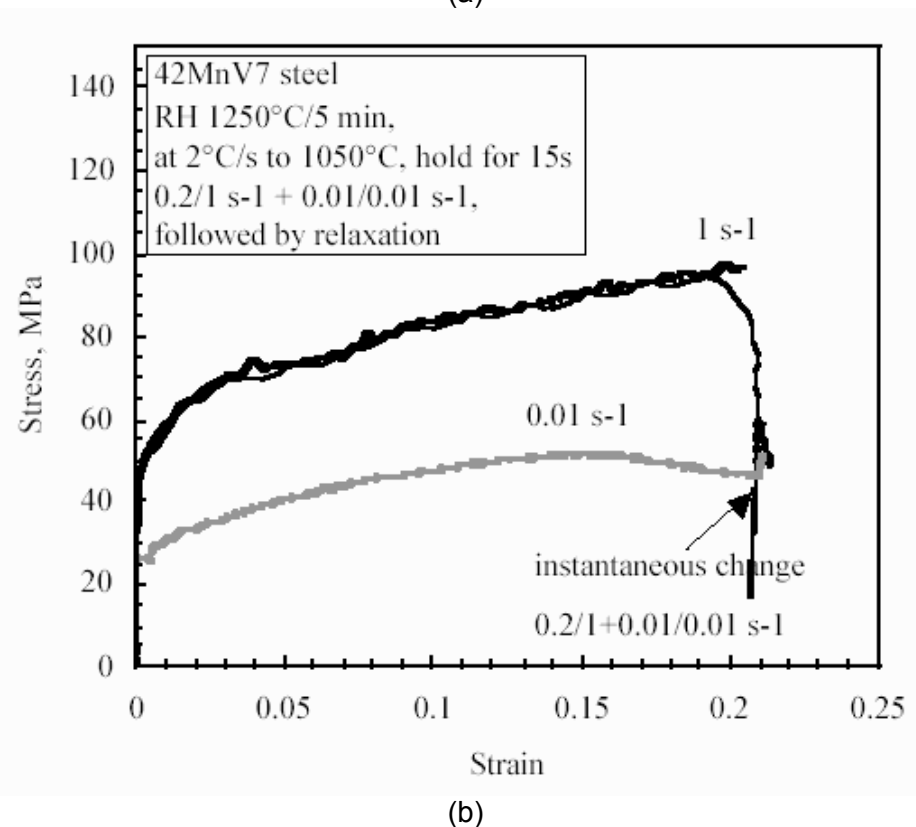
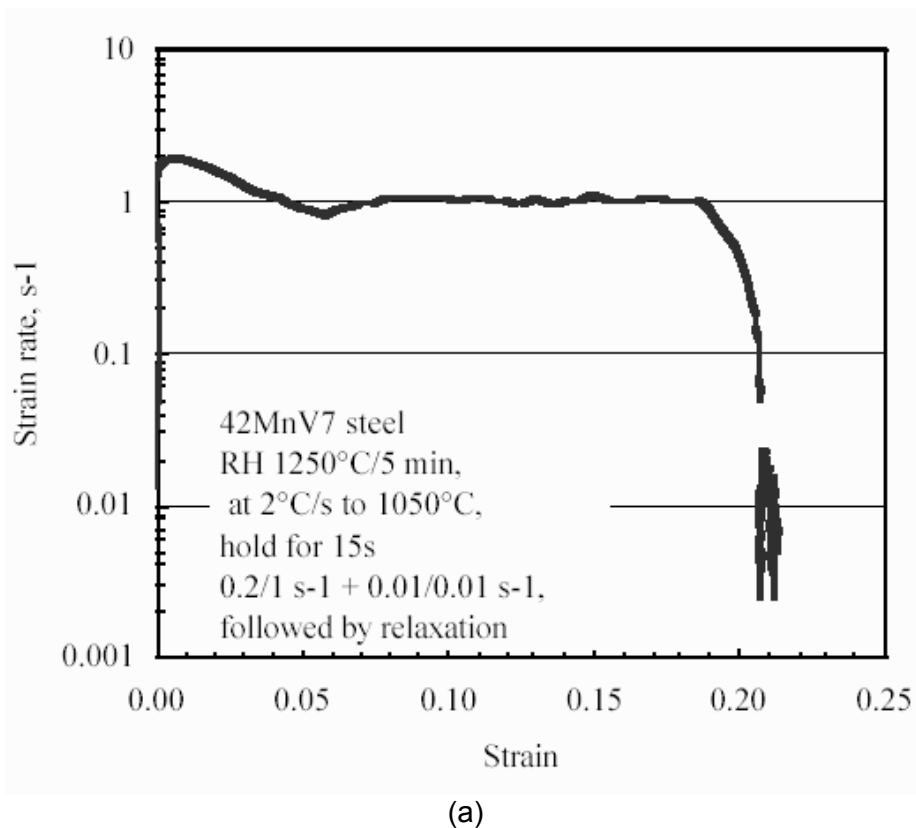
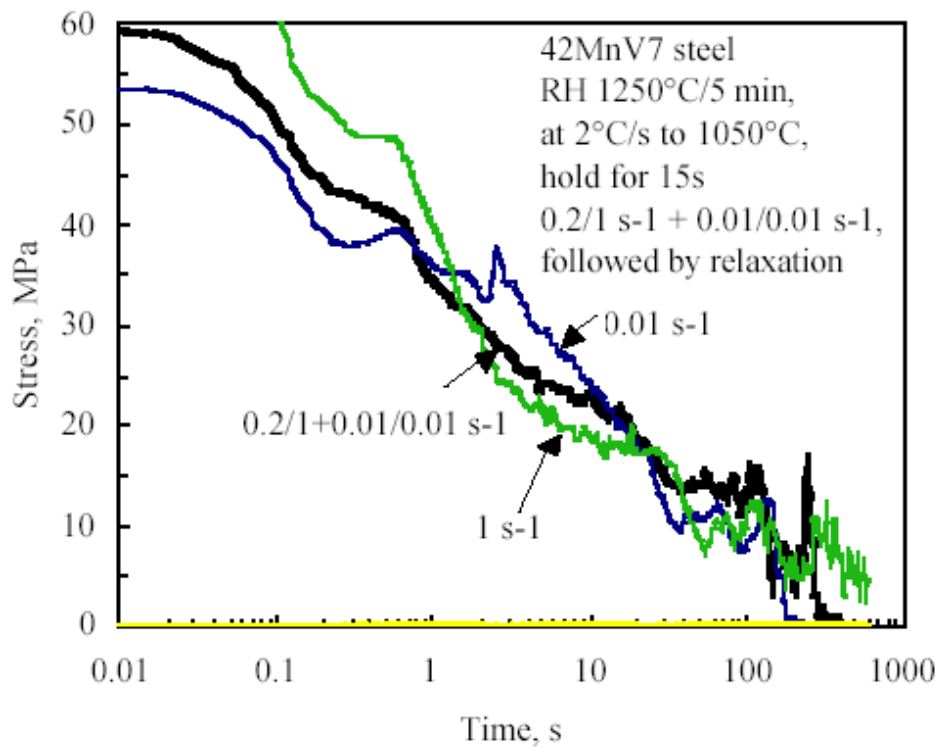
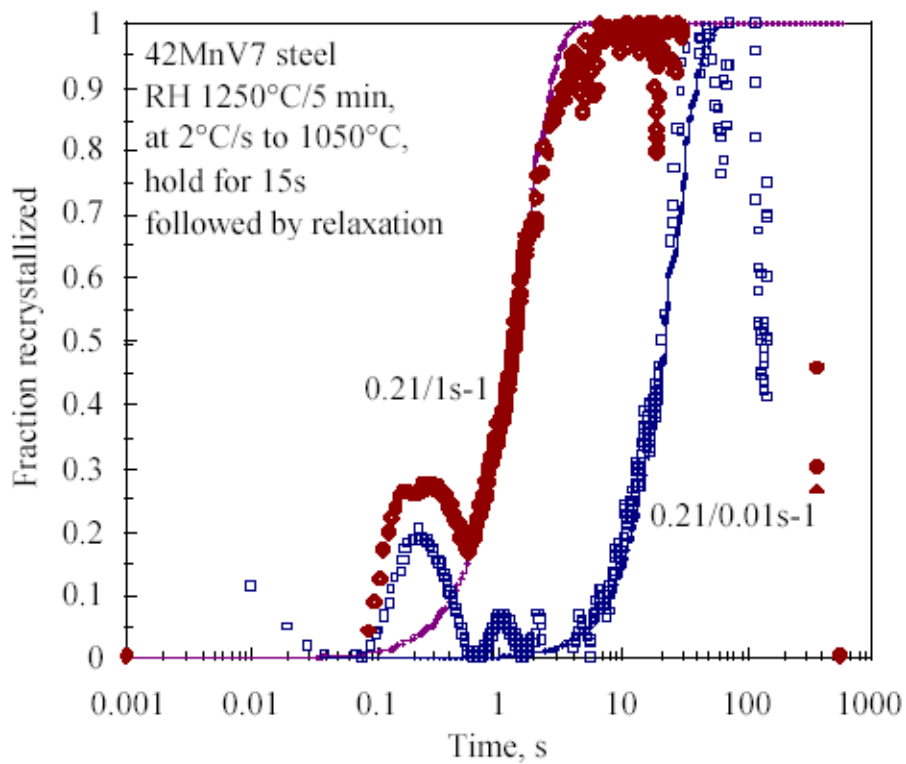


Fig. A6.9(a and b): Three strain rate histories (a) and their effects on flow stress (b).

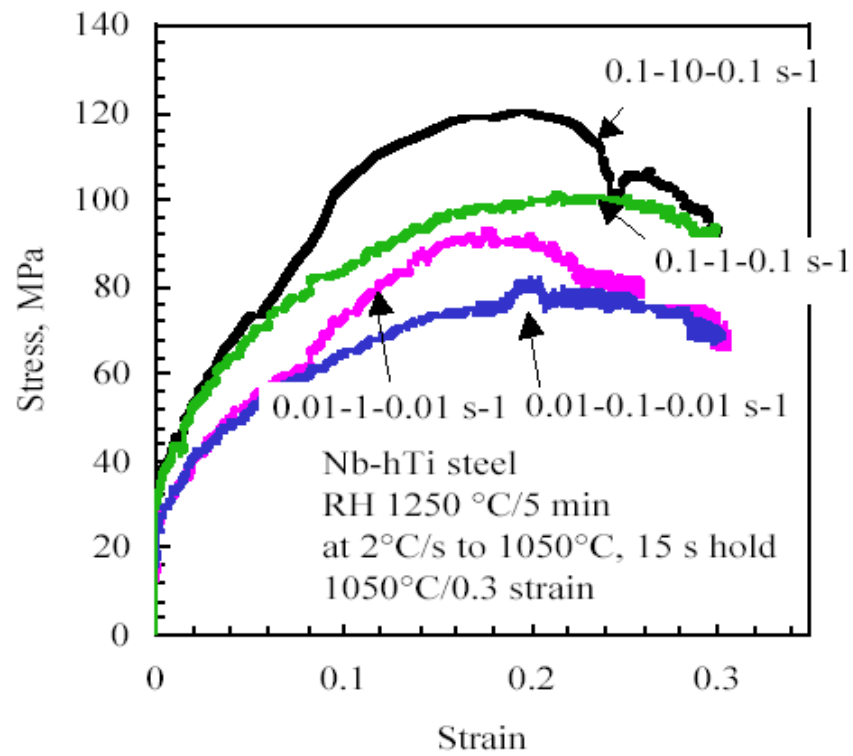


(a)

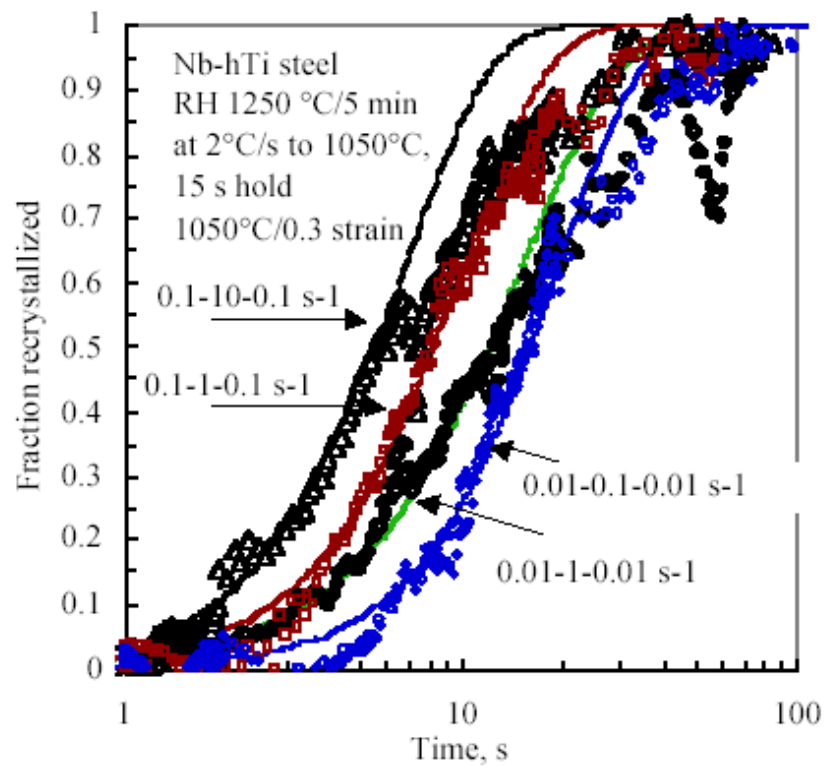


(b)

Fig. A6.10(a and b): Stress relaxation curves (a) and fractional softening curves (b). No softening following the strain rate change.

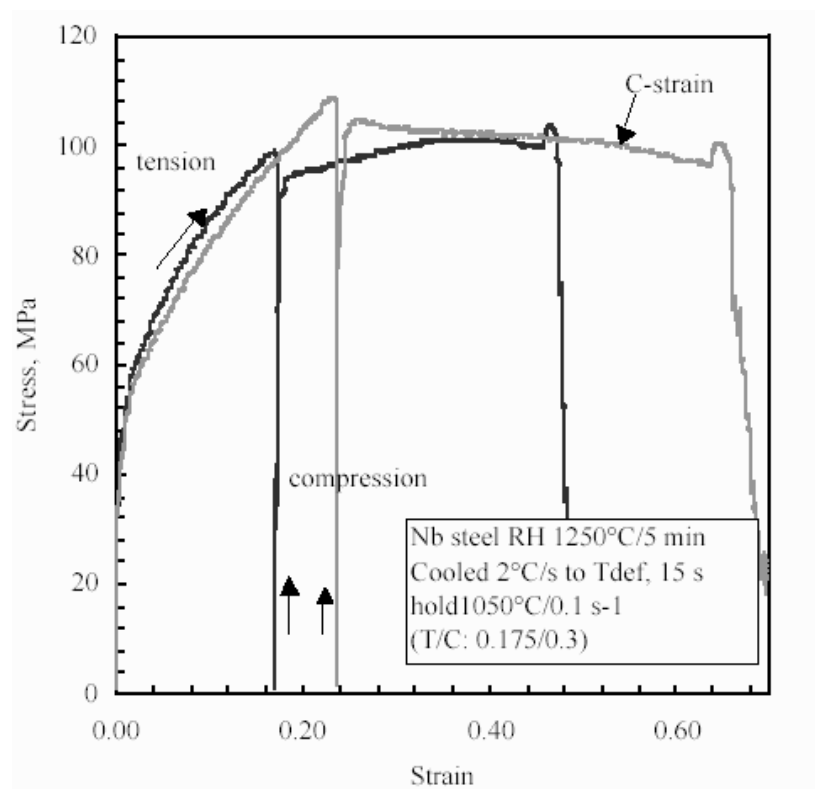


(a)

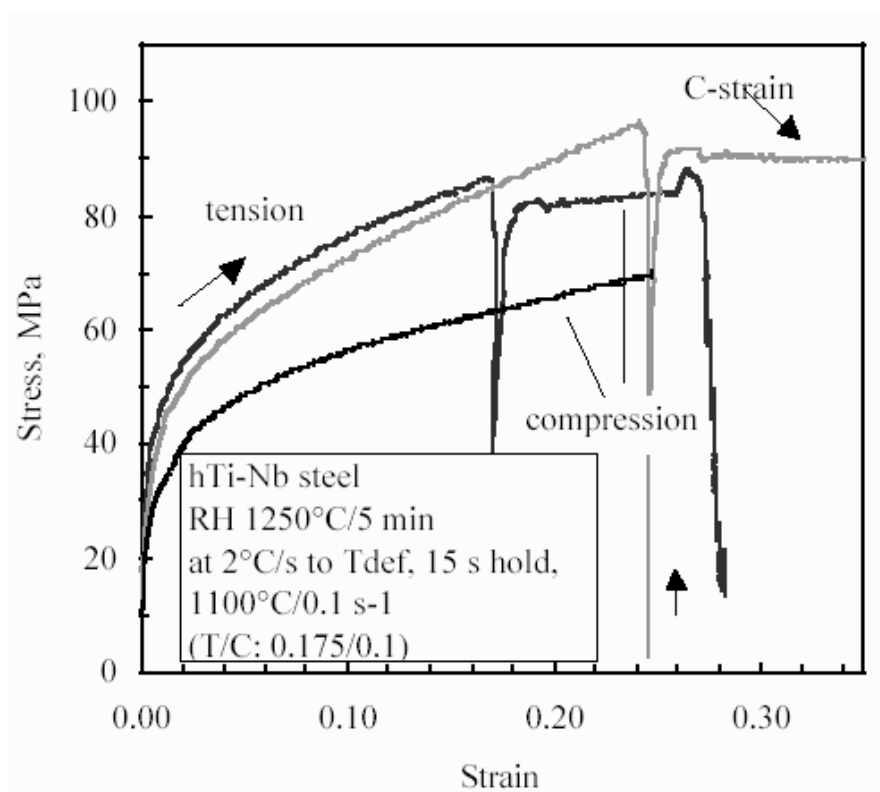


(b)

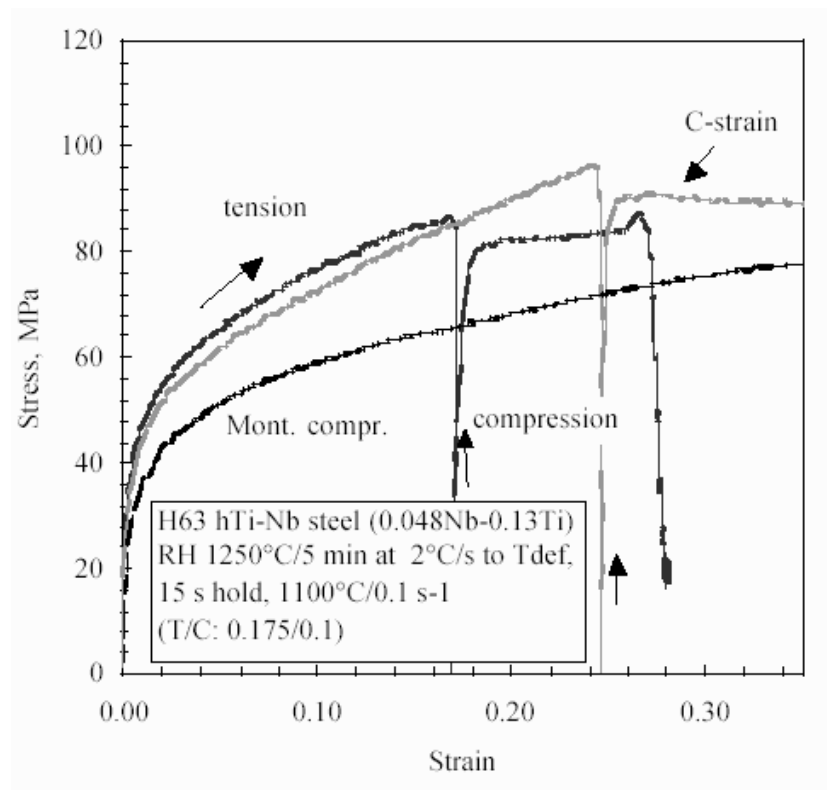
Fig. A6.11(a and b): Effect of continuous change in strain rate on flow stress (a) and fractional softening (b).



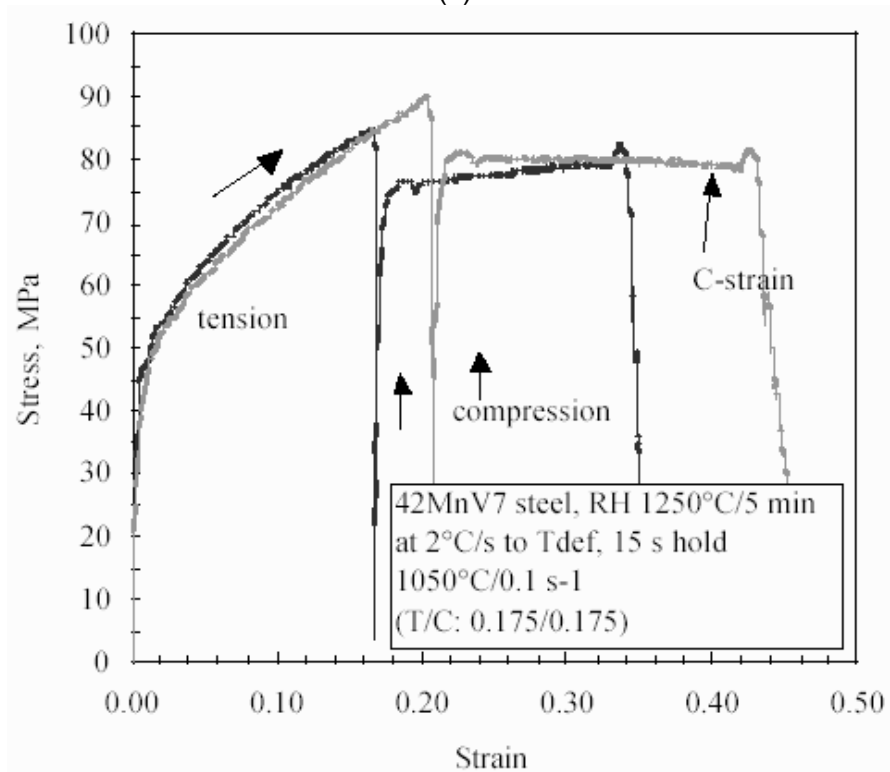
(a)



(b)



(c)



(d)

Fig. A6.12(a-d): Typical flow stress behaviours in reversal tests, changing from tension to compression at 0.175 nominal strain.

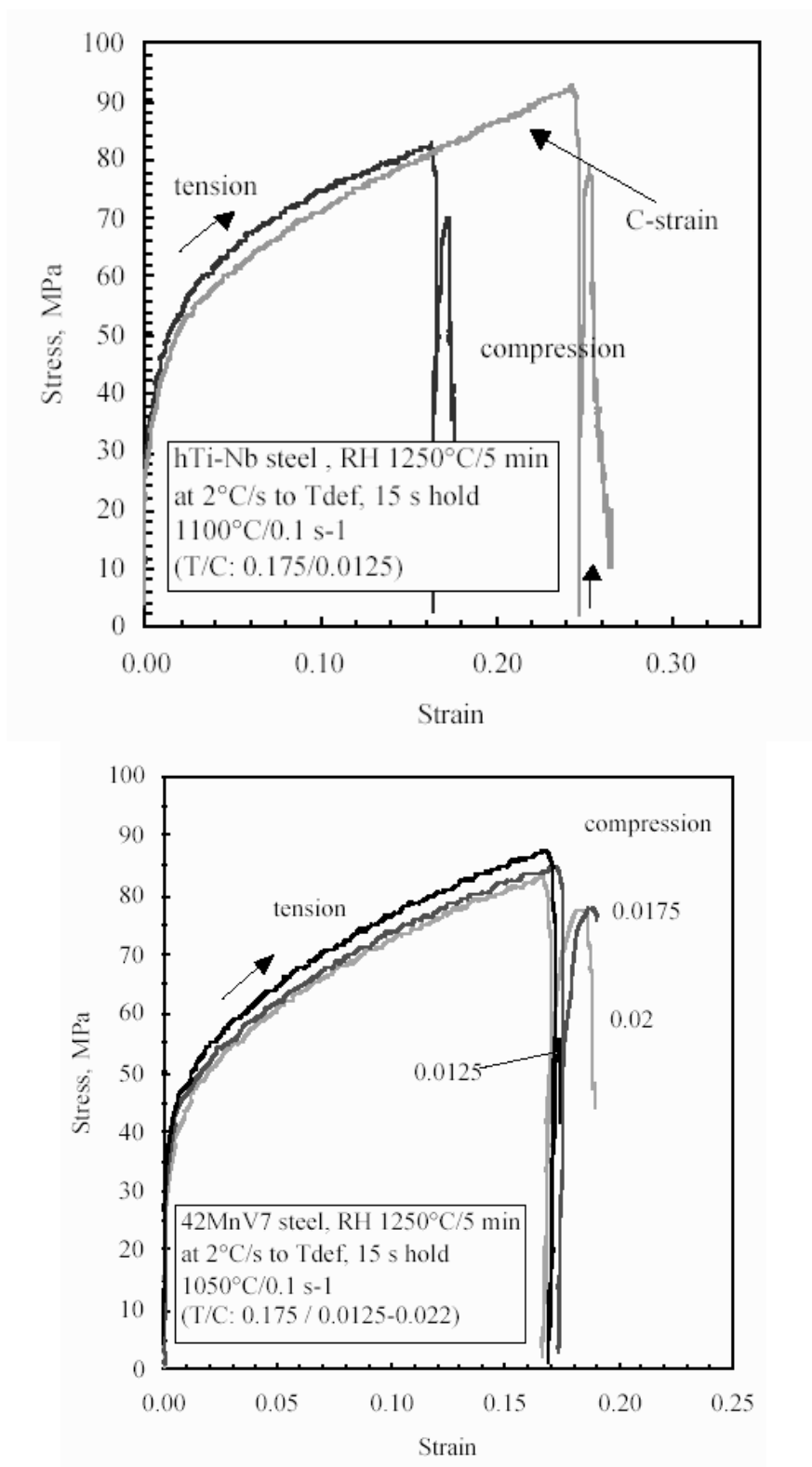
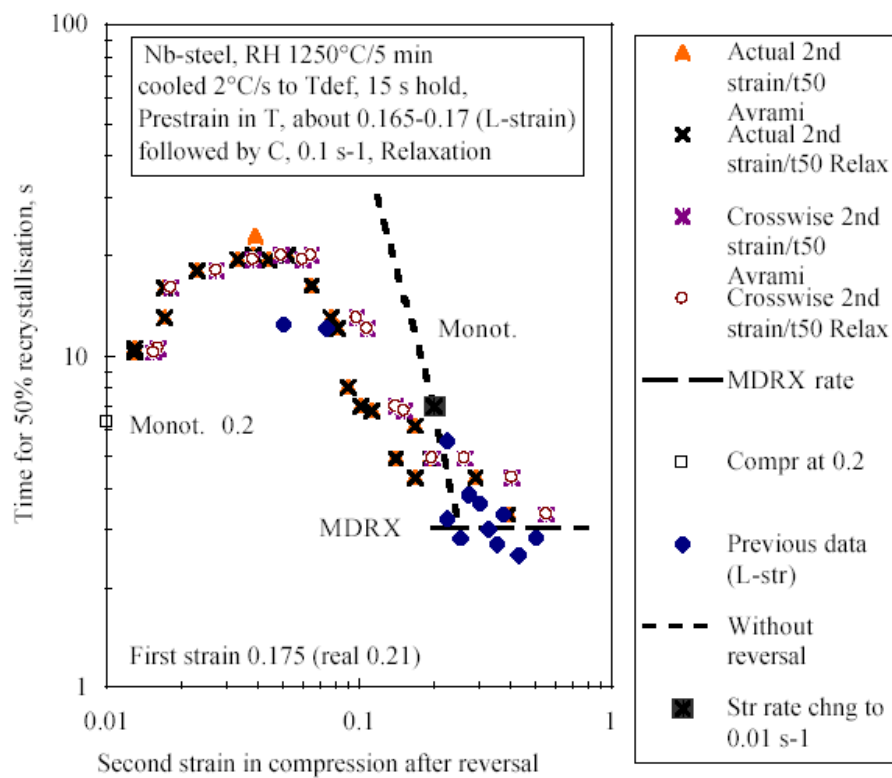
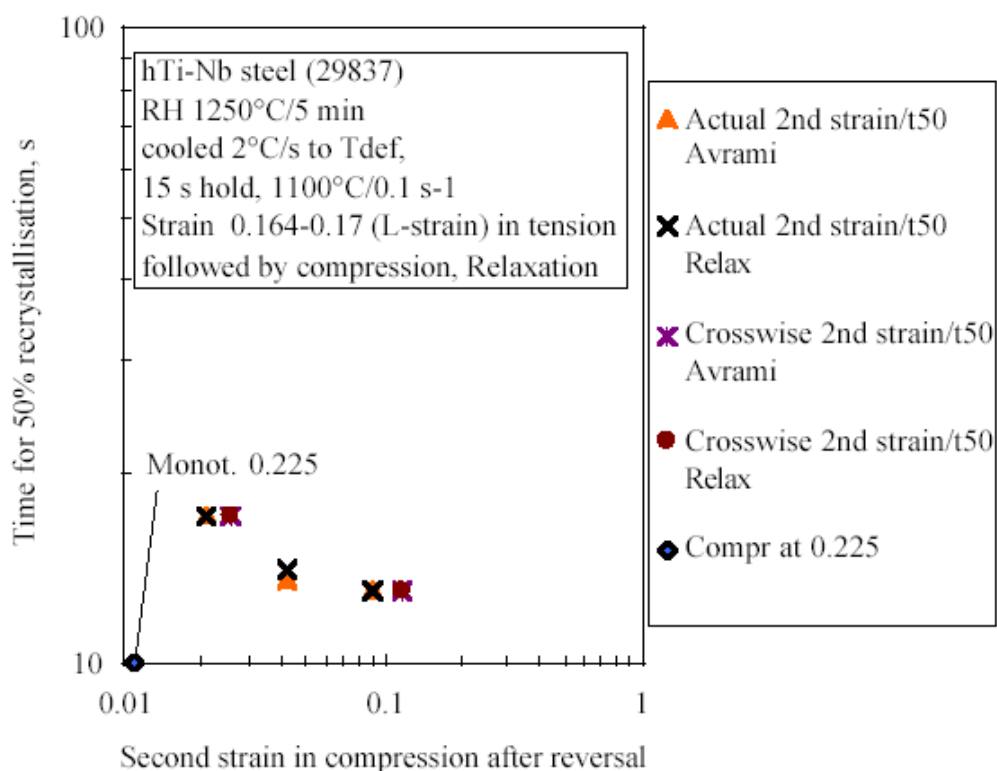


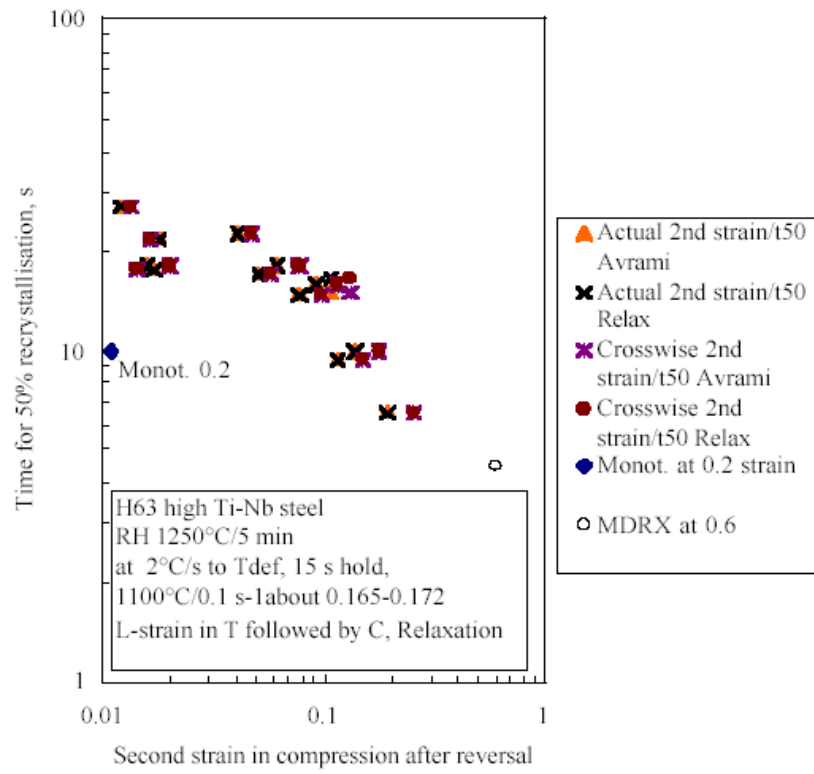
Fig. A6.13: Control of the second strain below 0.015 is hardly possible



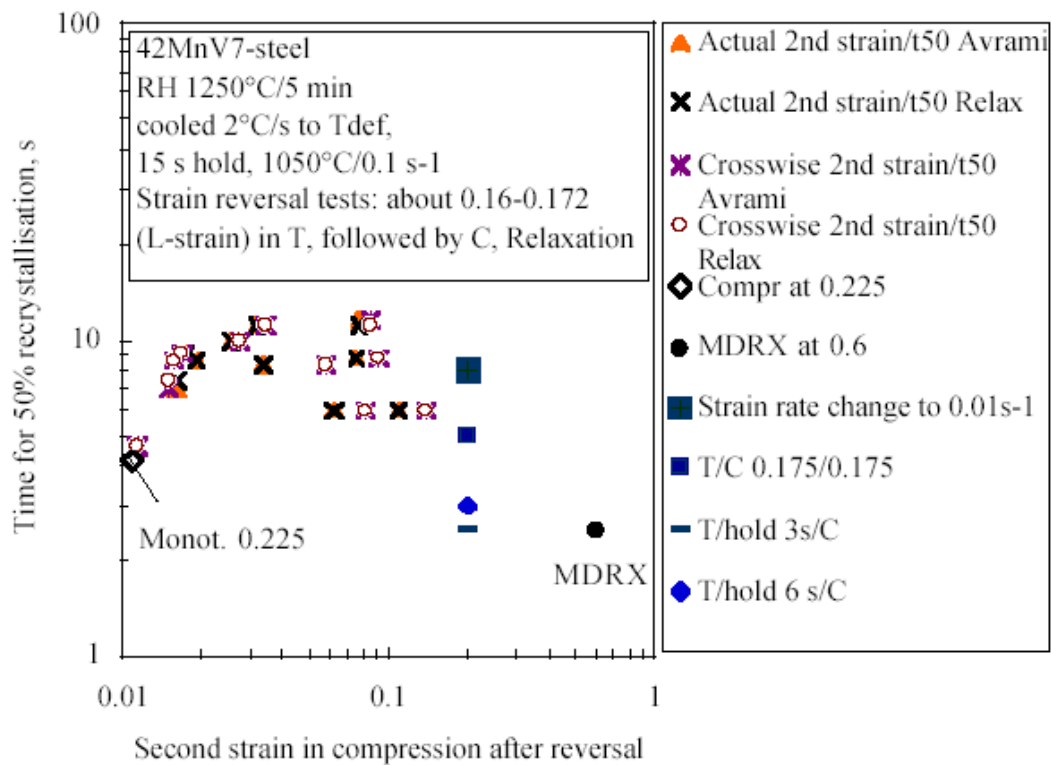
(a)



(b)



(c)



(d)

Fig. A6.14(a-d): Effect of the second compressive strain on t_{50} time after 0.175 nominal tensile strain in various steels. (a) Nb-steel (b) Nb-hti steel (c) another Nb-hti steel (d) 42MnV7 steel.

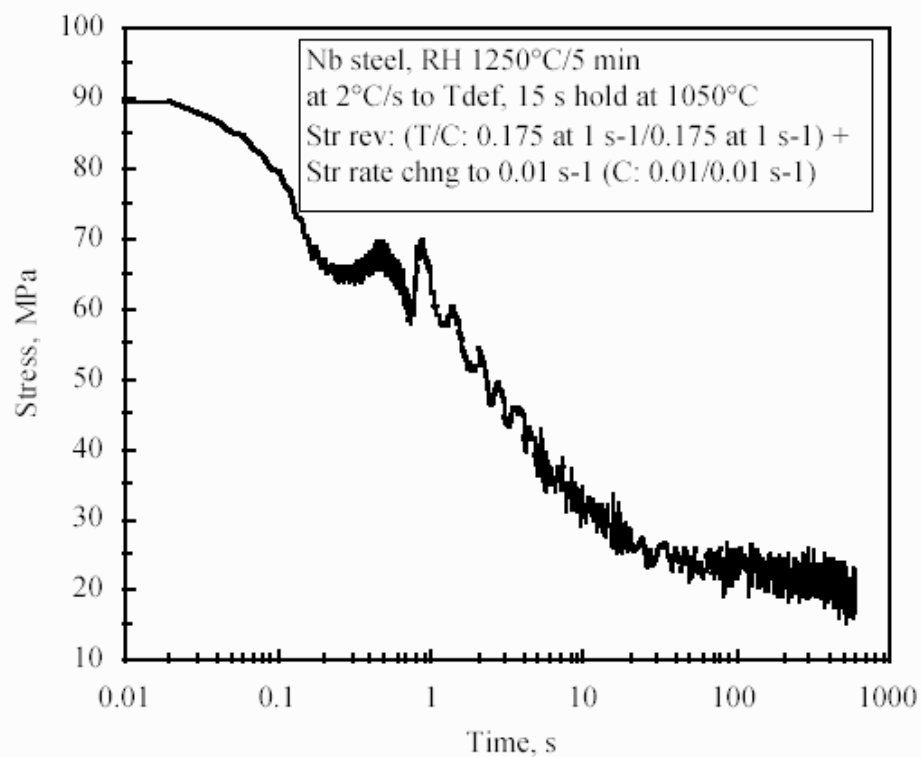
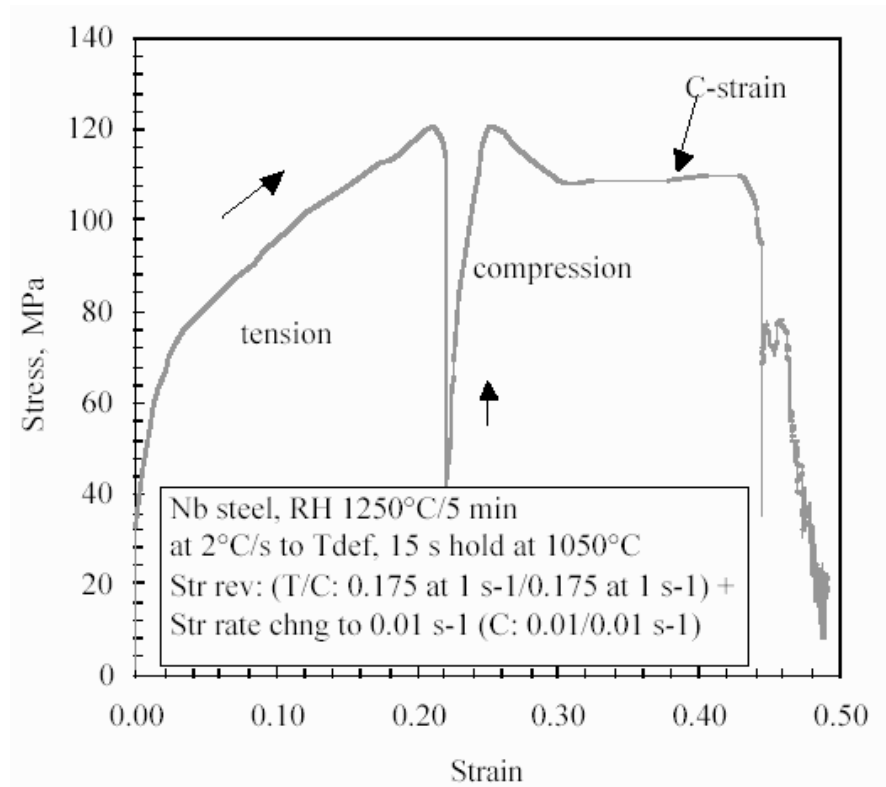


Fig. A6.15: Reversal/strain rate change test on the Nb steel, the flow stress behaviour and the corresponding relaxation curve.

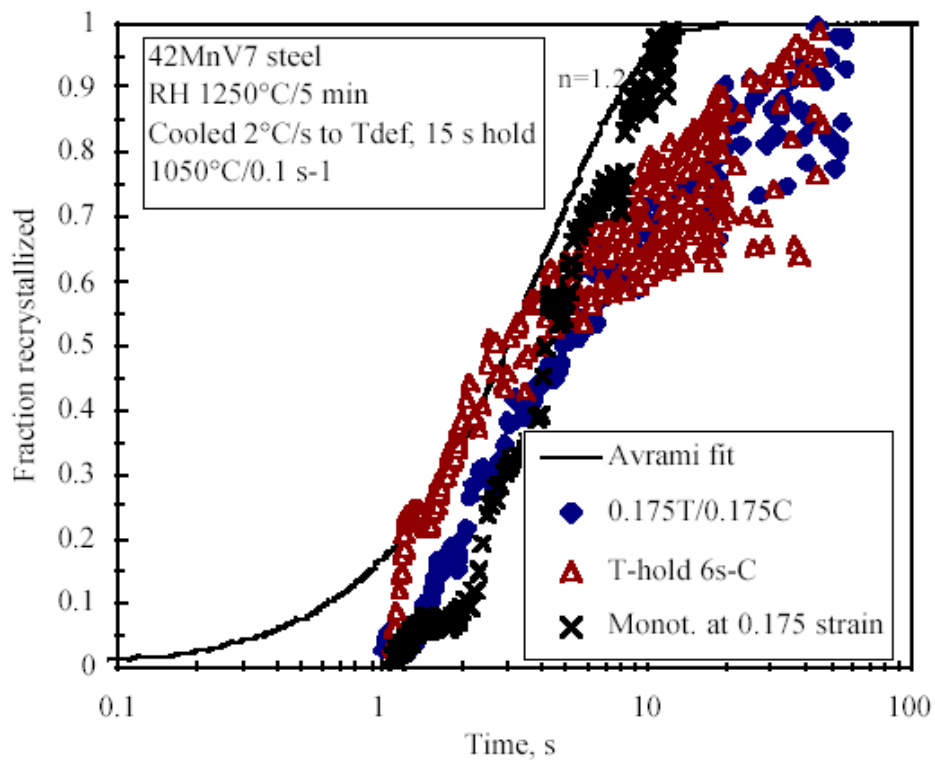
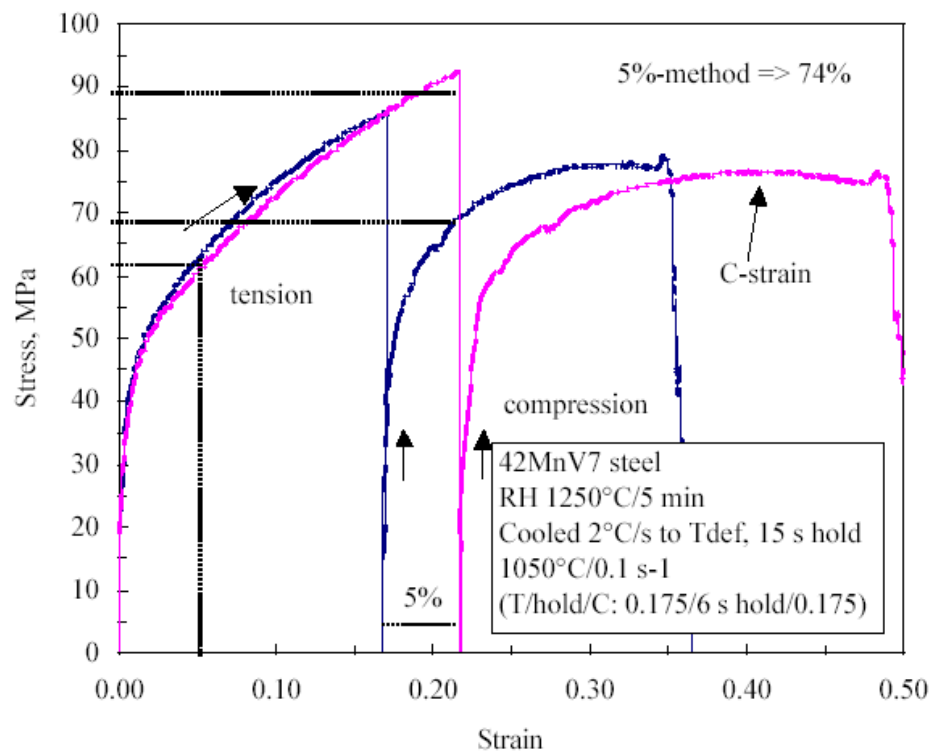
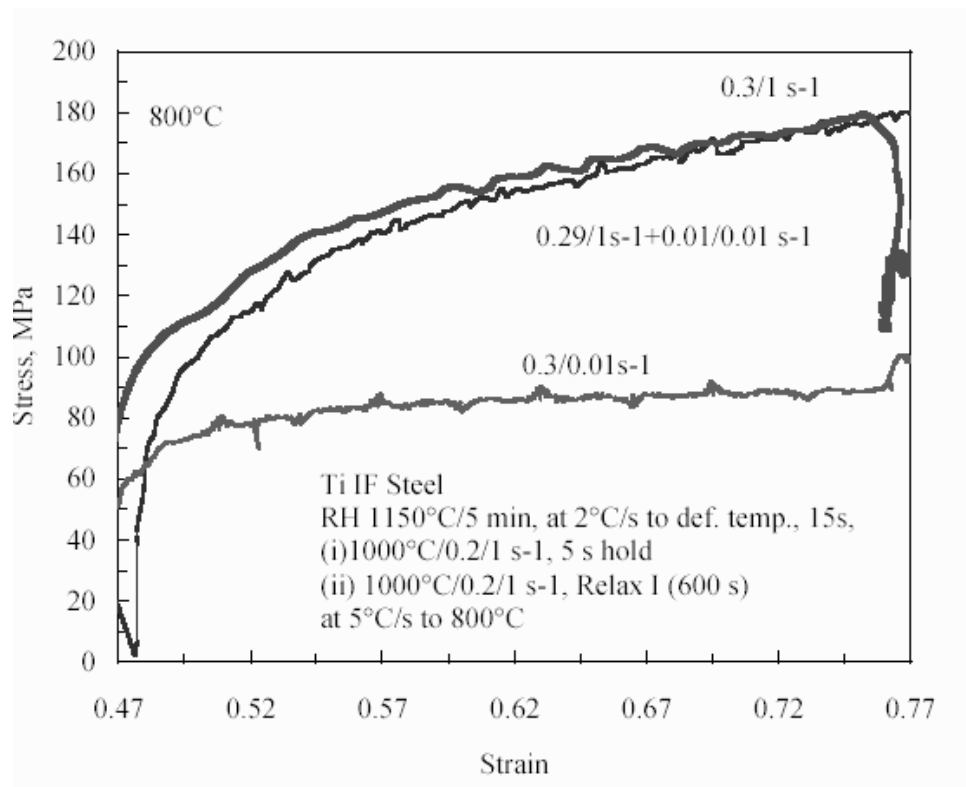
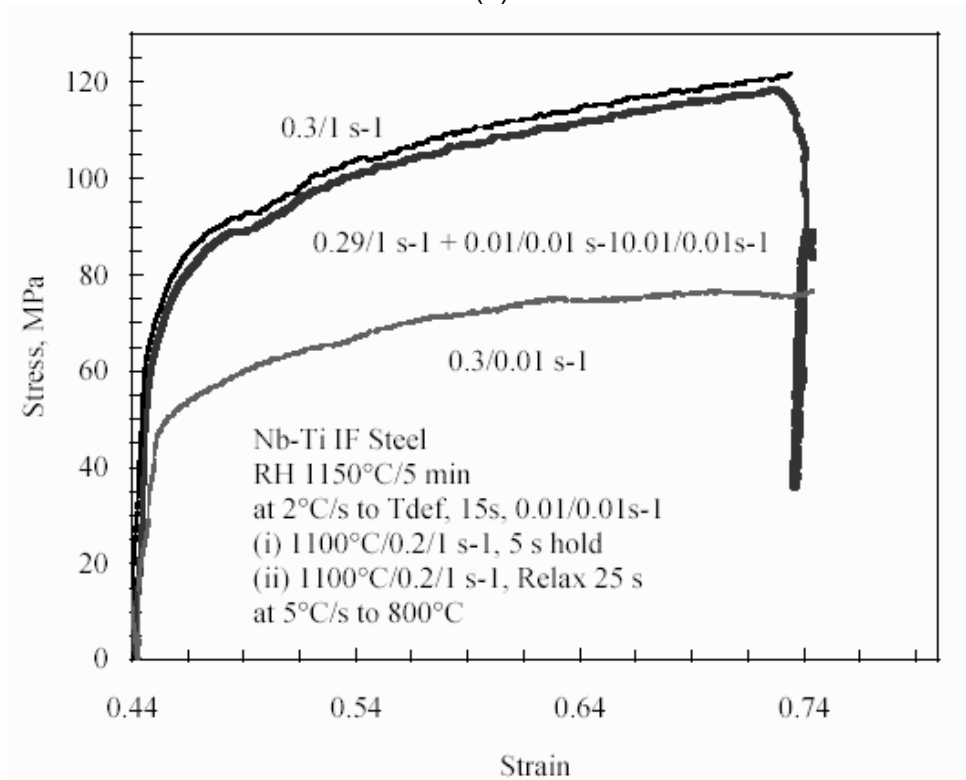


Fig. A6.16: Strain reversal t/c and t-holding for 6 s-c tests.

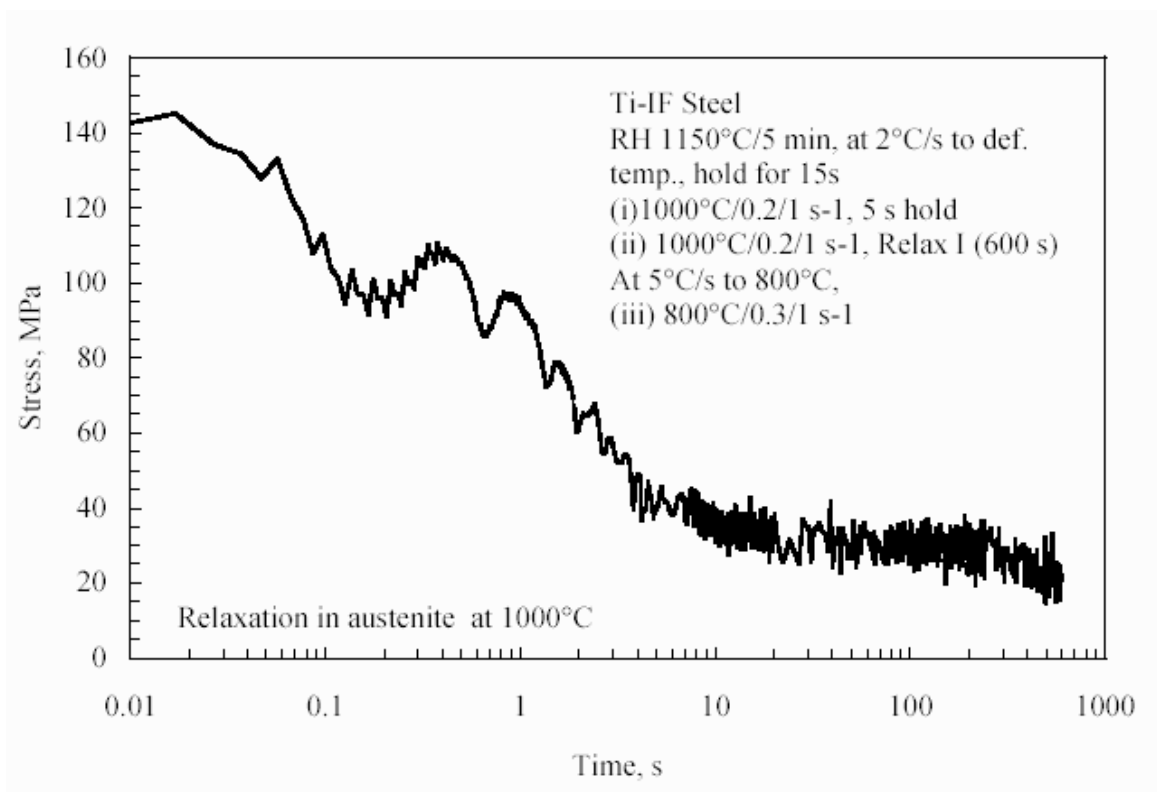


(a)

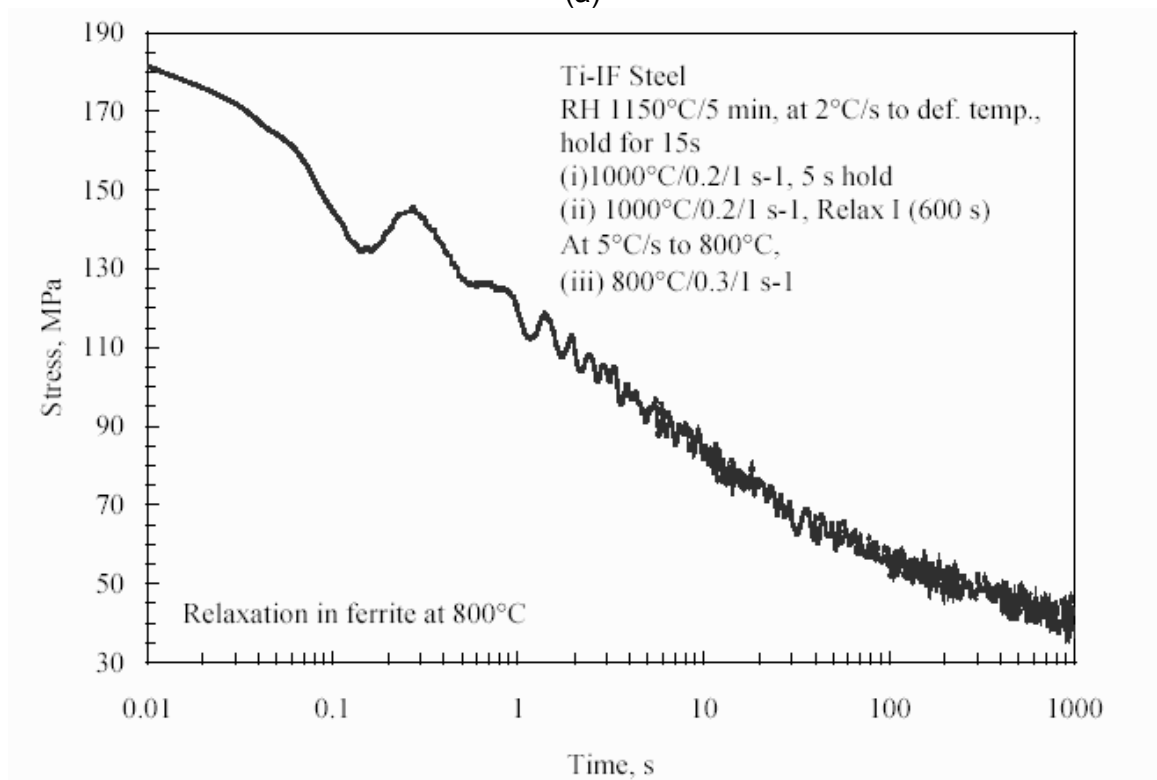


(b)

Fig. A6.17(a and b): Influence of strain rate history on flow stress of Ti IF steel at 800°C, in monotonic and strain rate change modes. Ti IF steel (a) and Nb-Ti IF steel (b).



(a)



(b)

Fig. A6.18(a and b): Comparison of typical relaxation curves for (a) austenite and (b) ferrite.

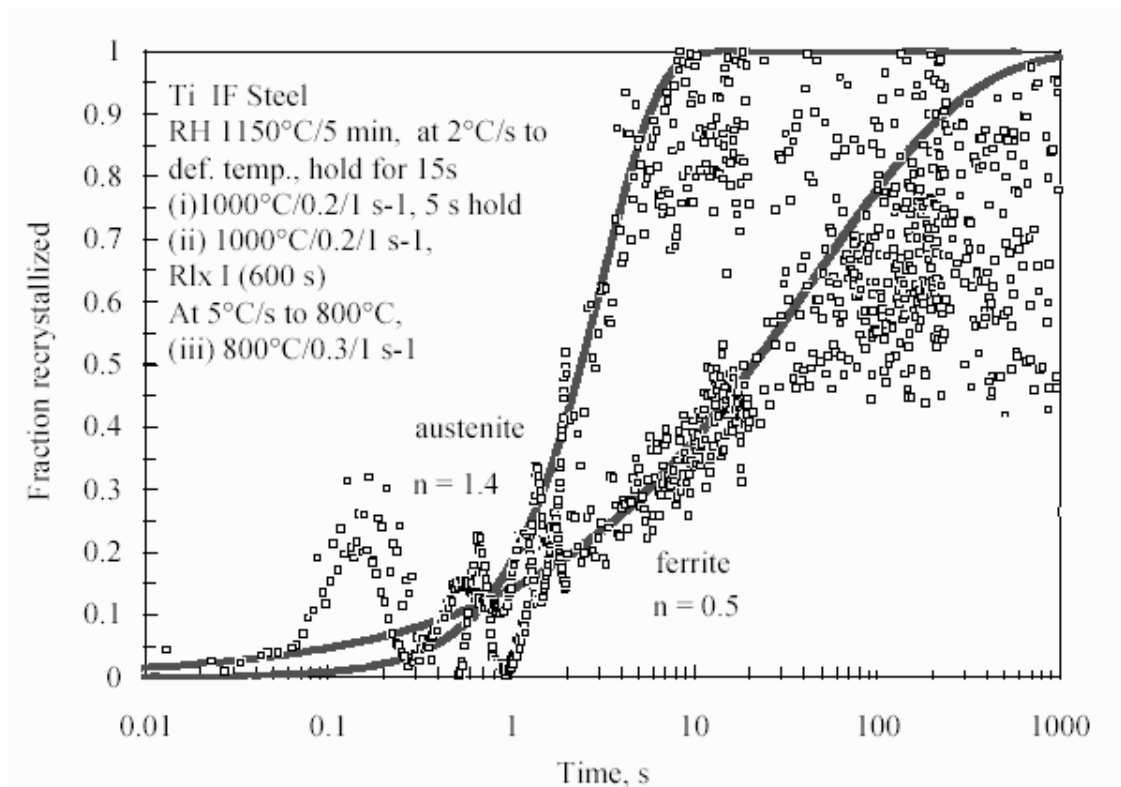


Fig. A6.19: Fractional softening curves for austenite and ferrite. Notice the different Avrami exponents.

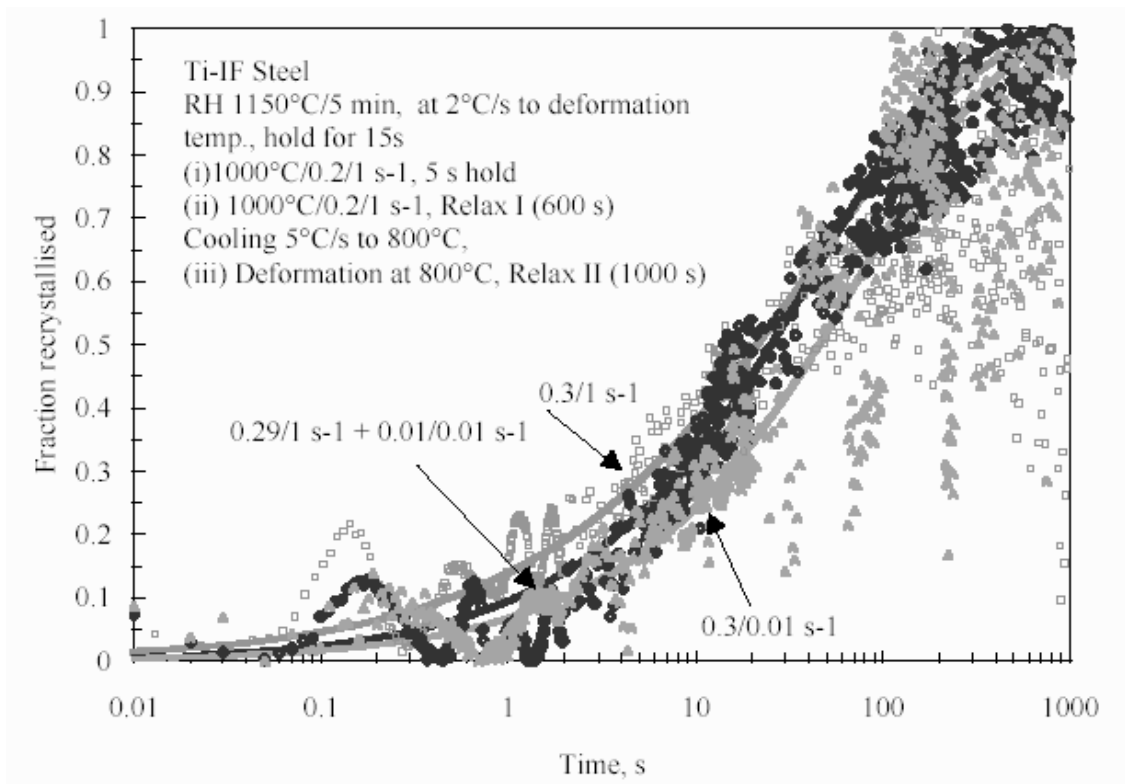


Fig. A6.20: Effect of strain rate history on fractional softening of ferrite in Ti-IF steel

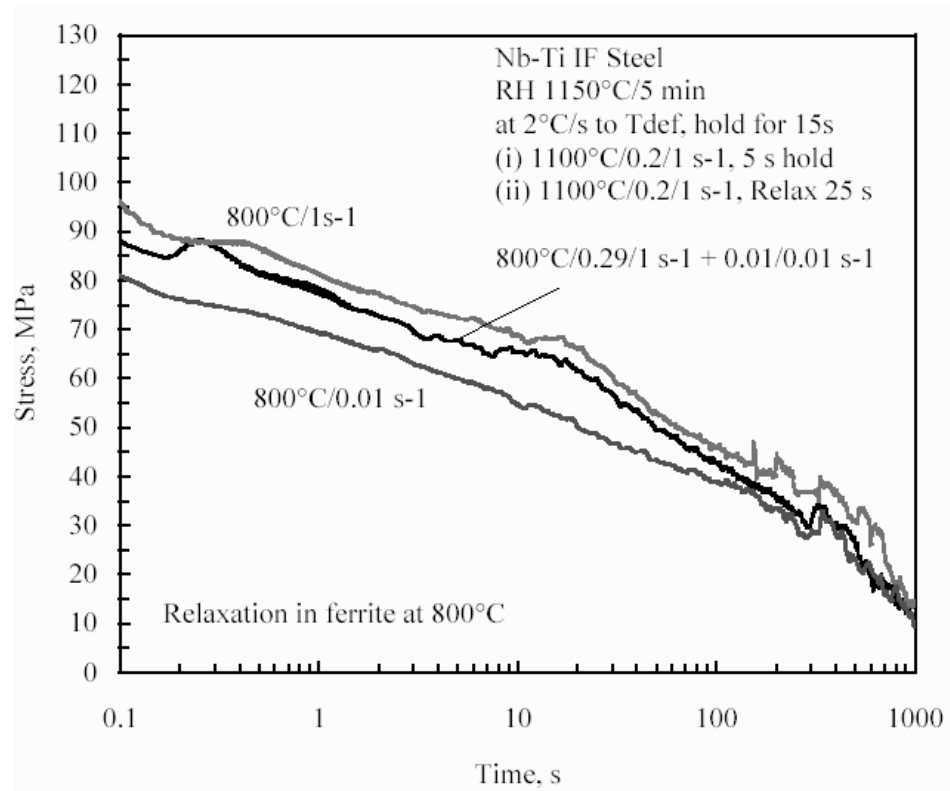
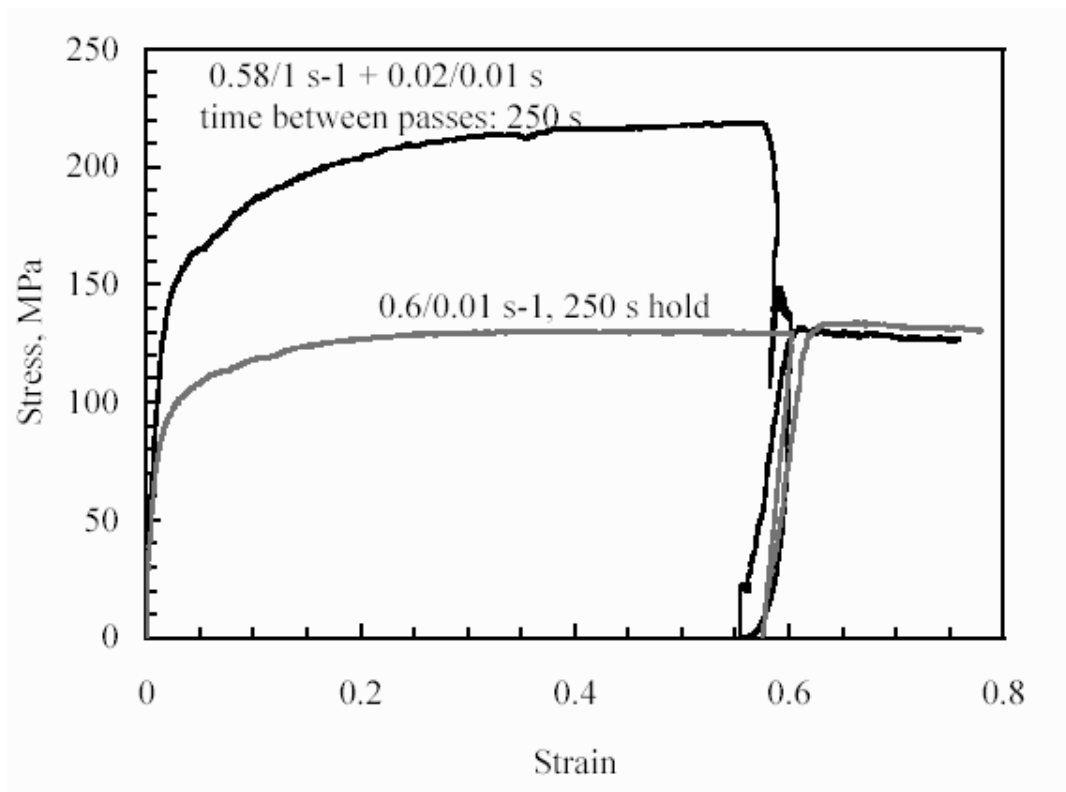
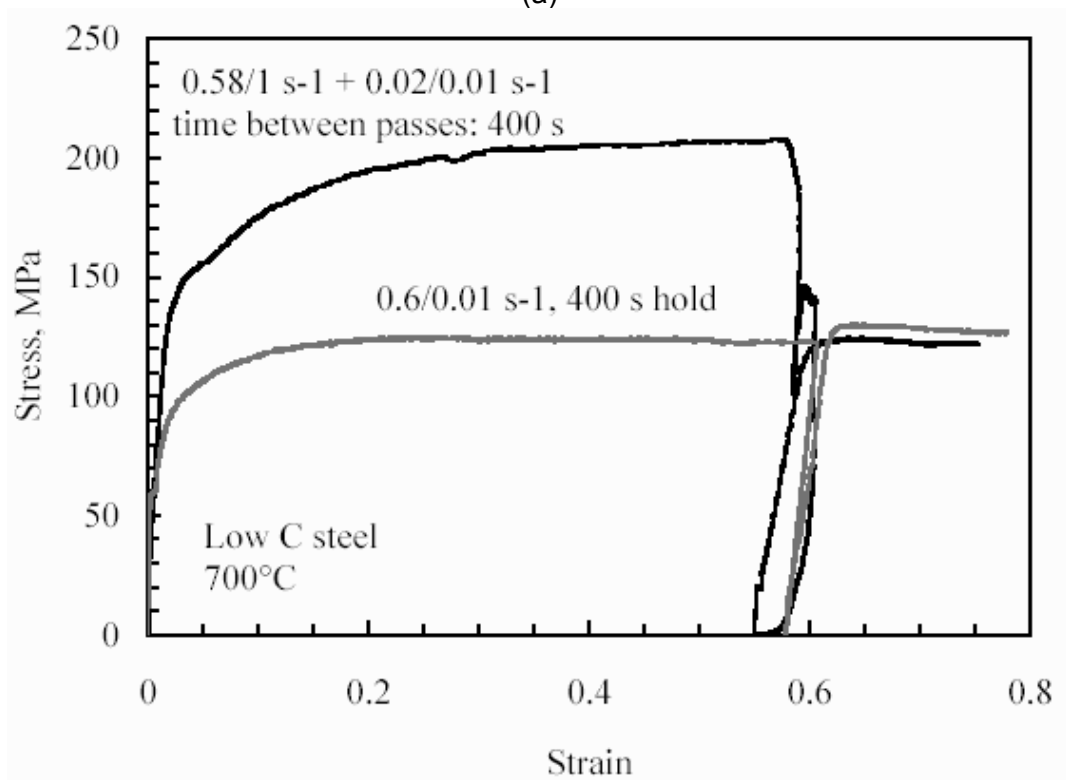


Fig. A6.21: Effect of strain rate history on relaxation curves at 800°C in Nb-Ti IF steel.



(a)



(b)

Fig. A6.22(a and b): Influence of the strain rate history on the flow stress of low carbon steel at 700°C. Holding for (a) 250 s (b) 400 s.

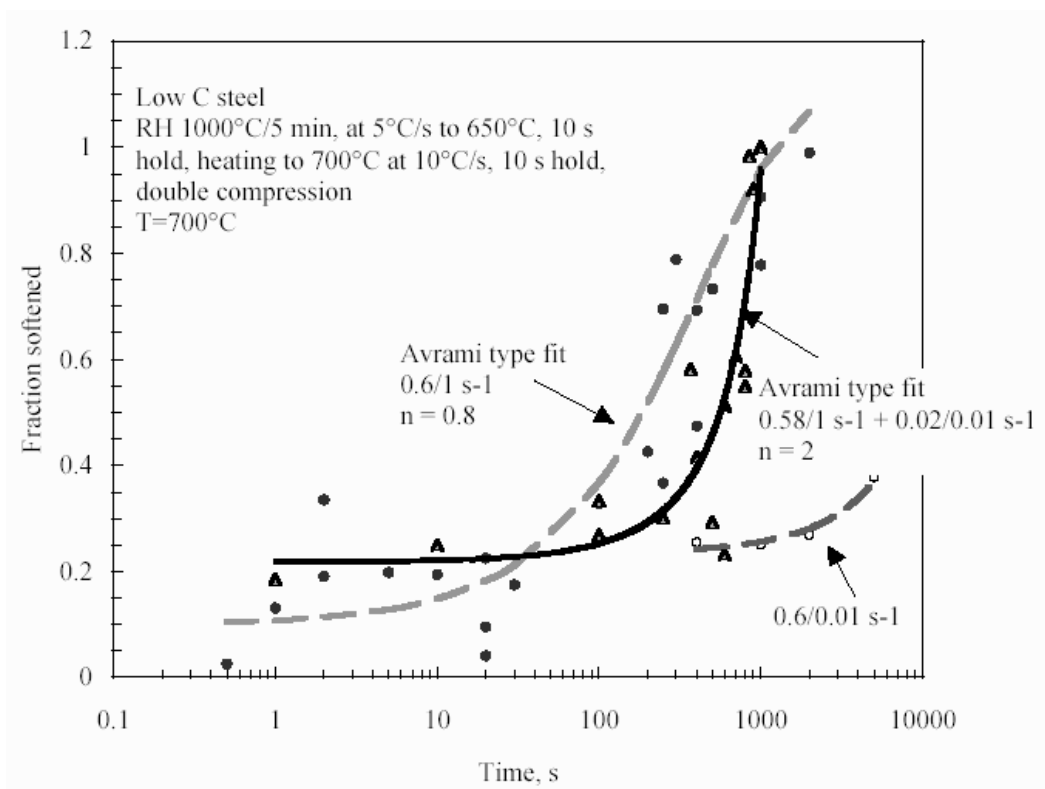


Fig. A6.23: Recrystallisation of ferrite after 0.6 strain at 700°C.
Constant strain rate or strain rate change tests.

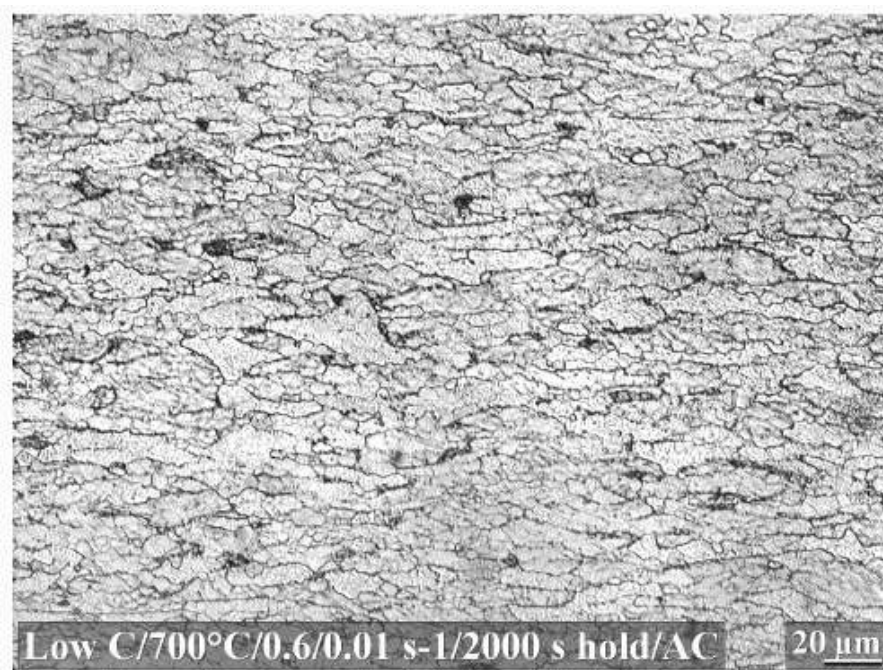
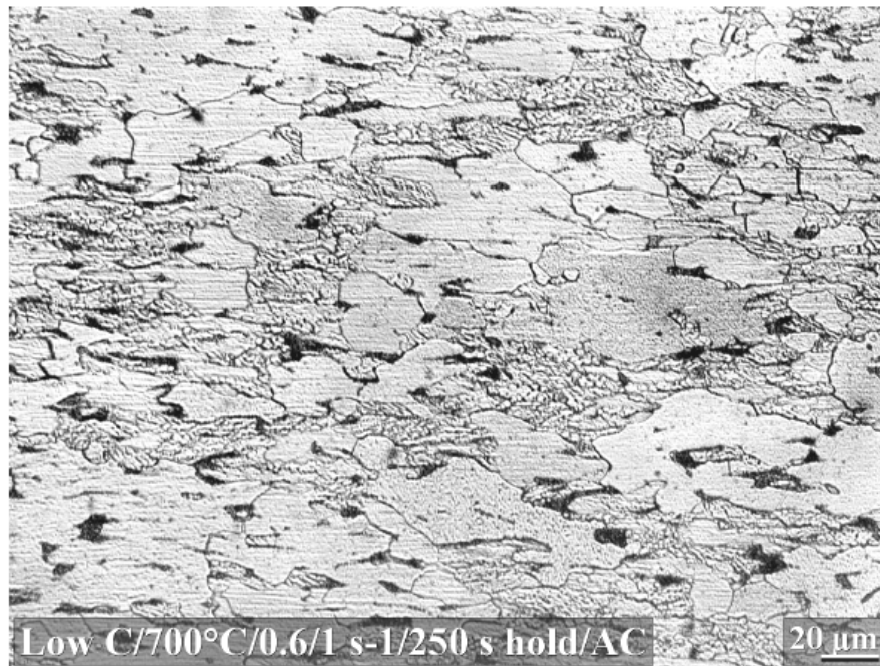
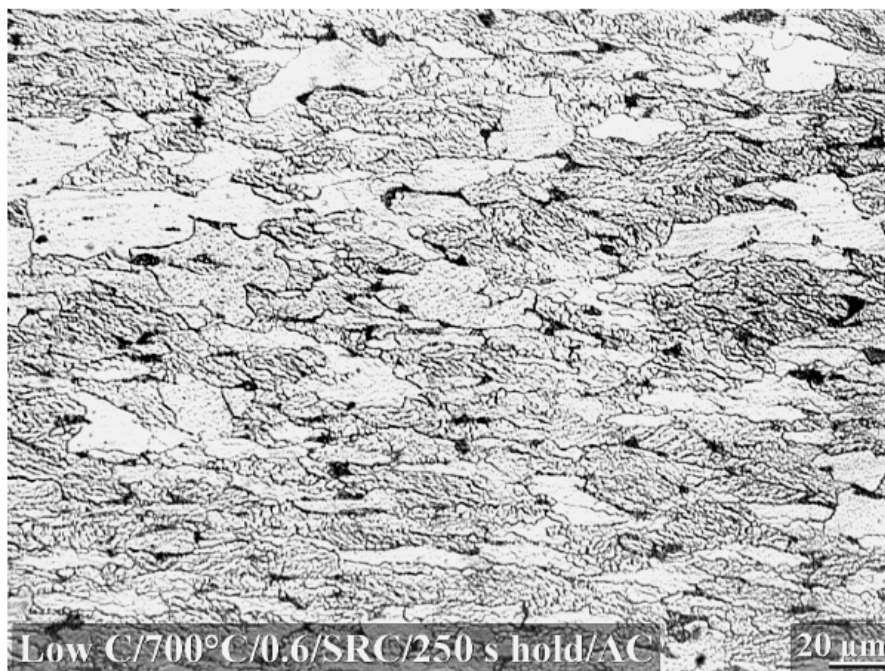


Fig. A6.24: Recrystallisation of ferrite after 0.6 strain at 700°C in 2000 s. Constant strain rate
of 0.01 s⁻¹.

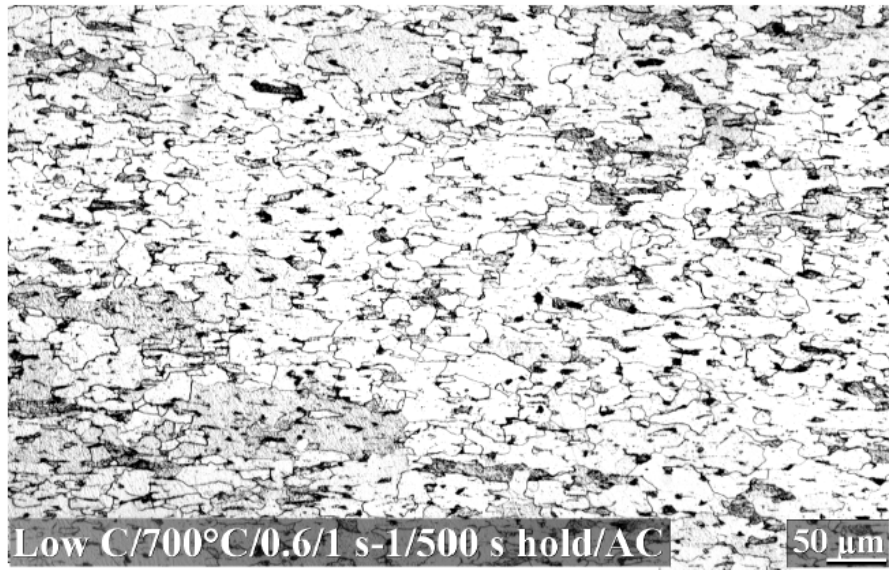


(a)

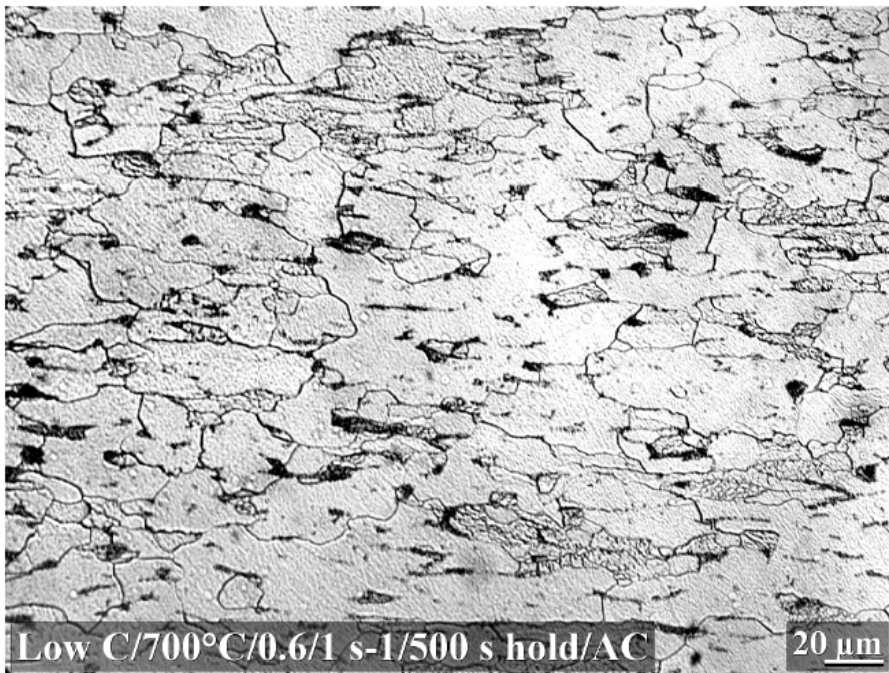


(b)

Fig. A6.25(a and b): Recrystallisation at 700°C in 250 s (a) at 1 s^{-1} (b) strain rate change

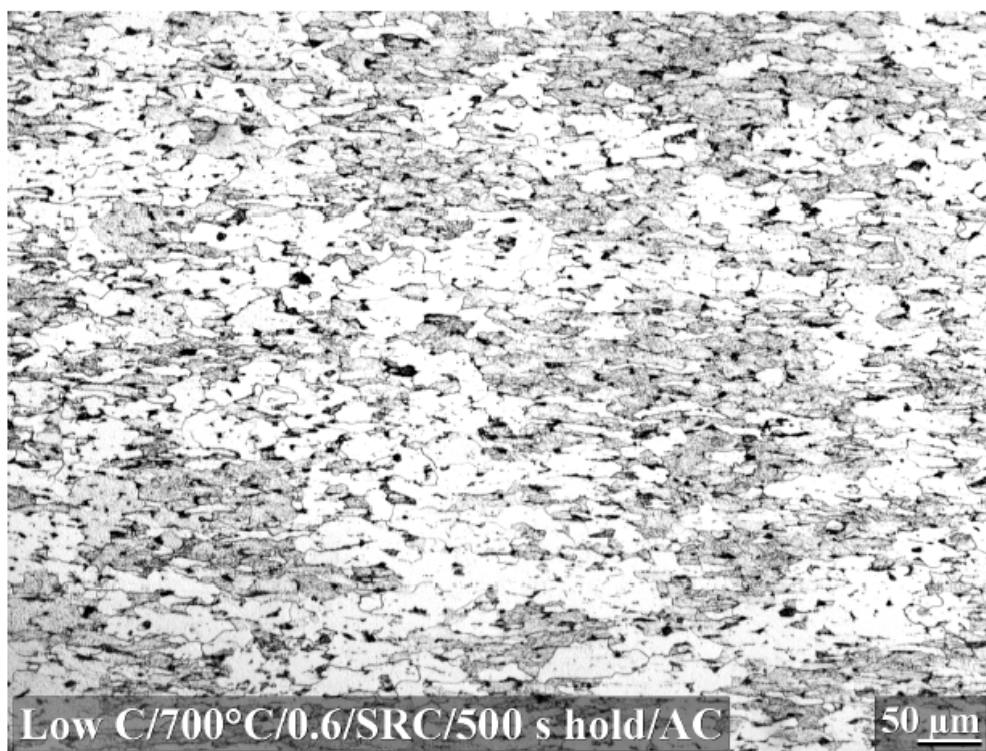


(a)

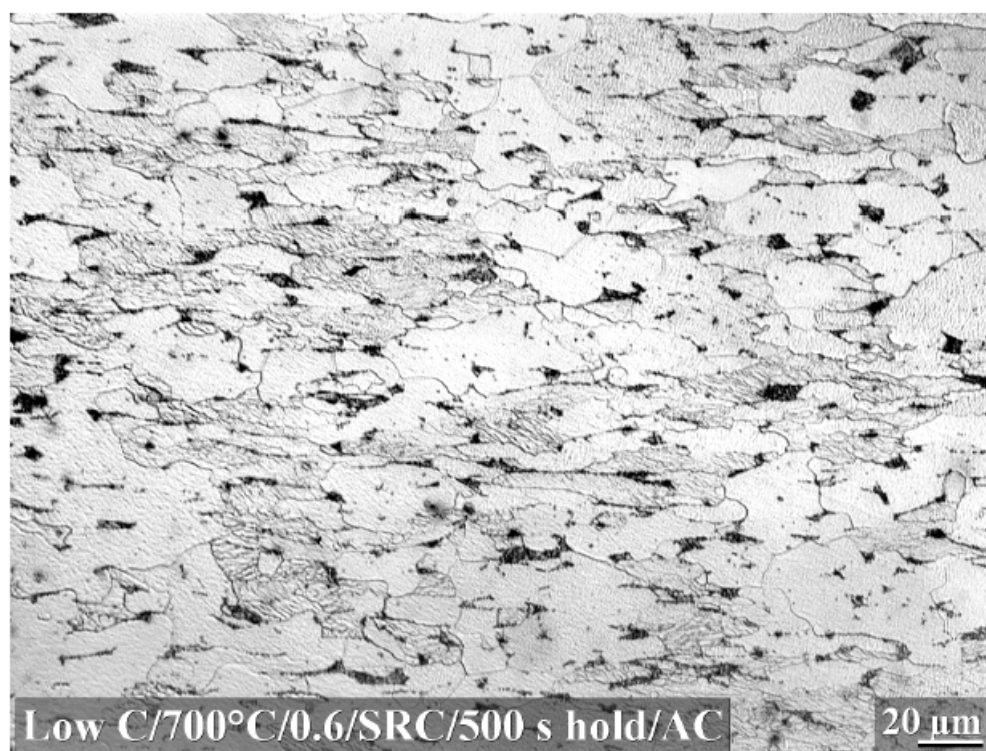


(b)

Fig. A6.26(a and b): recrystallisation at 700°C in 500 s after 0.6 strain at 1 s⁻¹, (a) lower magnification (b) higher magnification.



(a)



(b)

Fig. A6.27(a and b): Recrystallisation at 700°C in 500 s in strain rate change deformation (a) lower magnification (b) higher magnification.

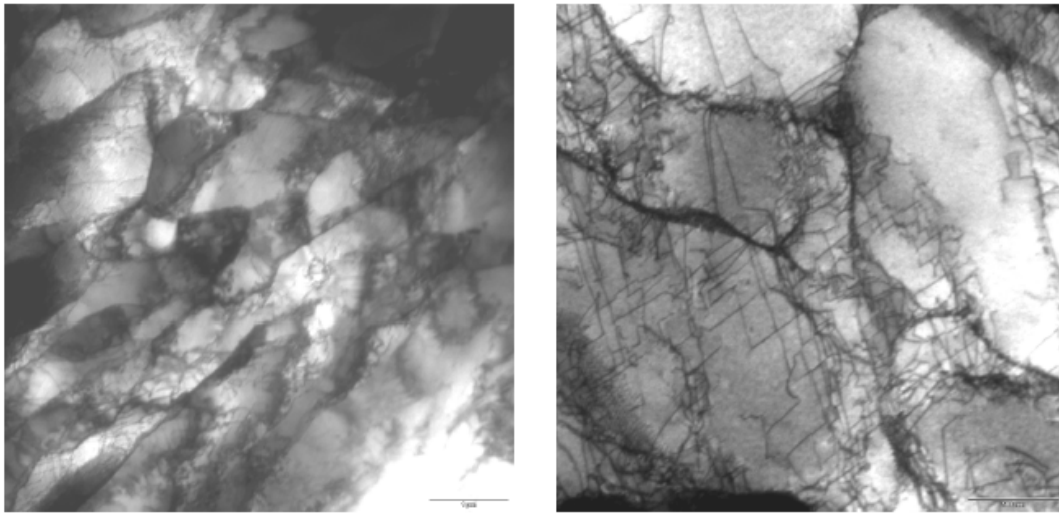


Fig. A6.28: Cell structure seen in TEM in the Ti-IF steel after straining at the constant strain rate of 1 s^{-1} .

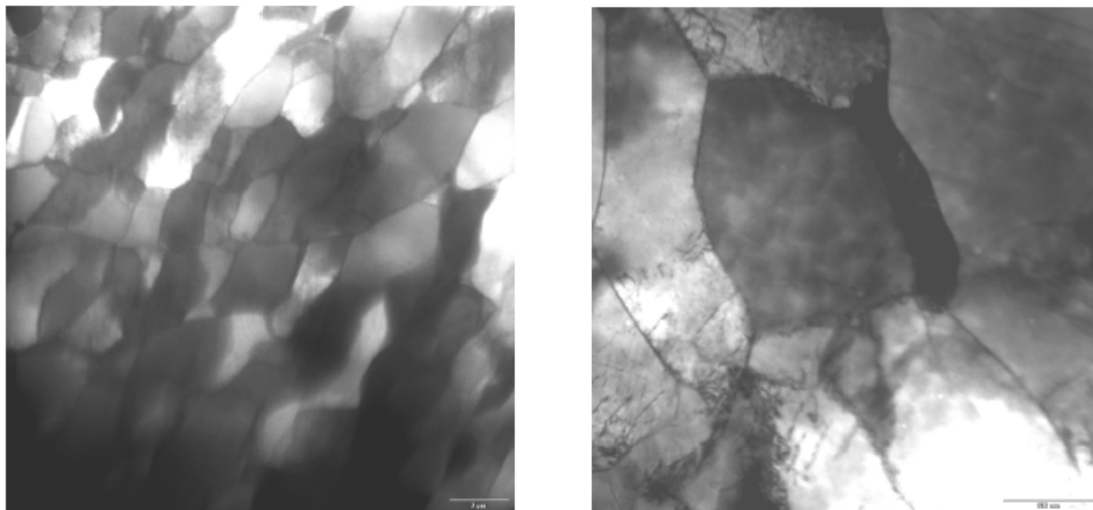


Fig. A6.29: Cell structure seen in TEM in Ti-IF steel after the strain rate change

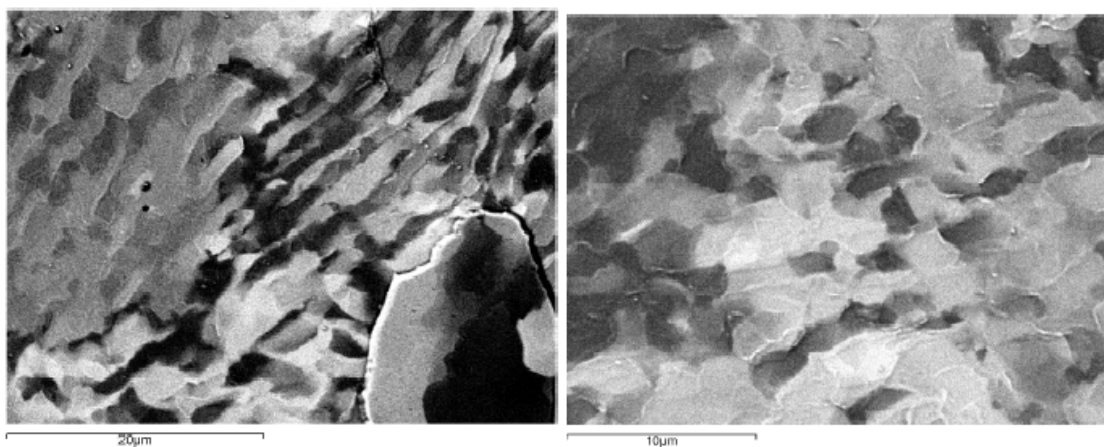


Fig. A6.30: Cells on the surface of Ti-IF steel strained to 0.3 strain at the constant strain rate of 1 s^{-1}

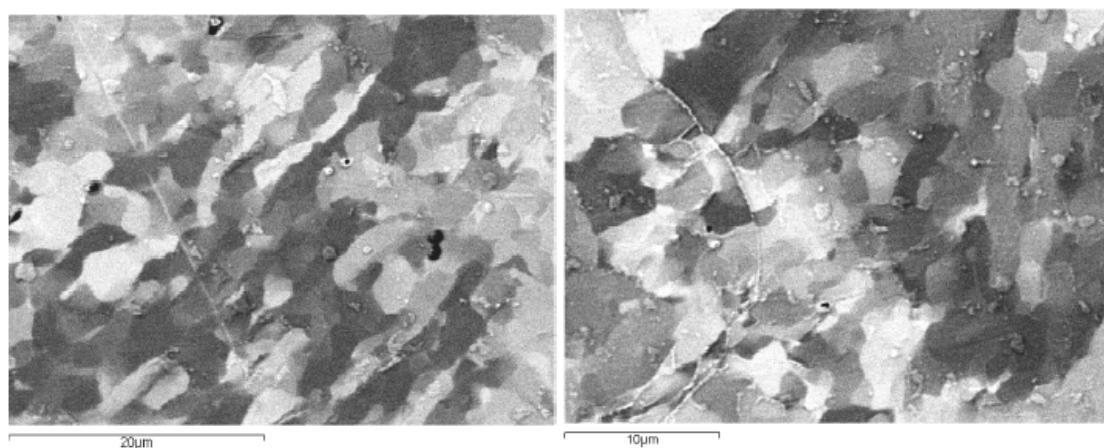


Fig. A6.31: Cells on the surface of Ti-IF steel strained to 0.27 strain at the changed strain rate

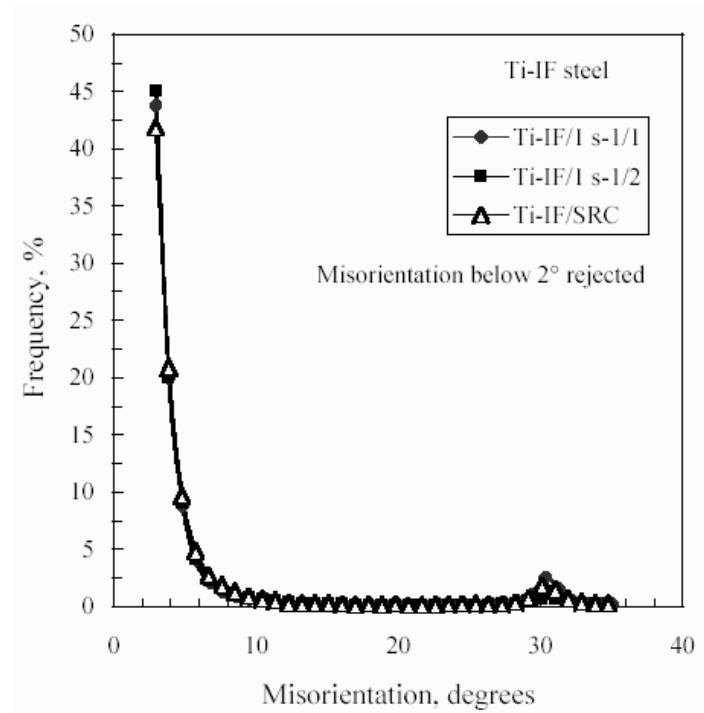


Fig. A6.32: Misorientation distribution of the grain boundaries in Ti-IF steel after constant and changed strain rate tests

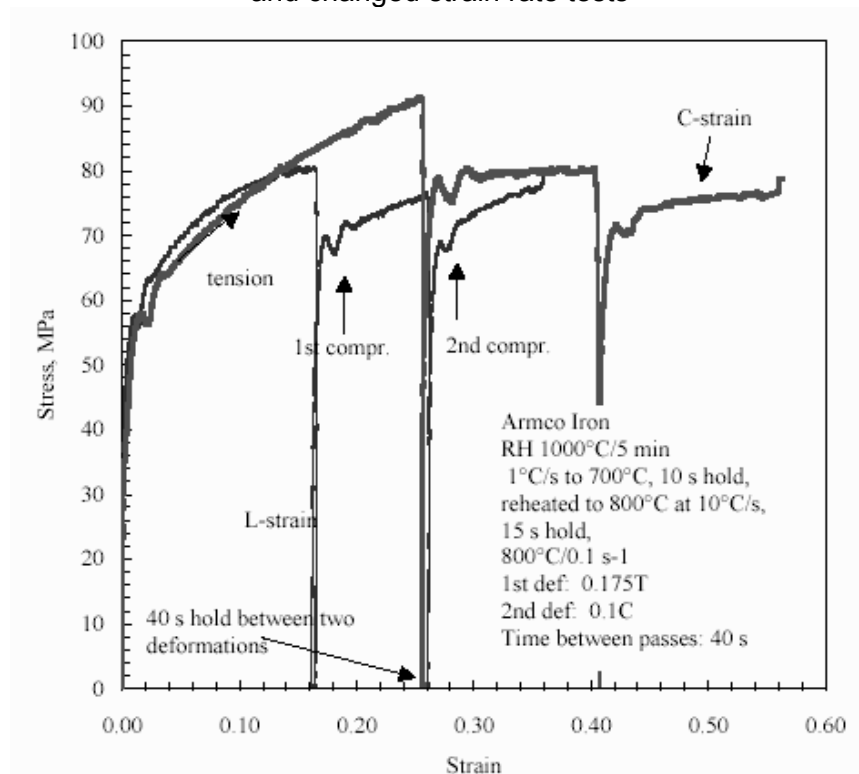
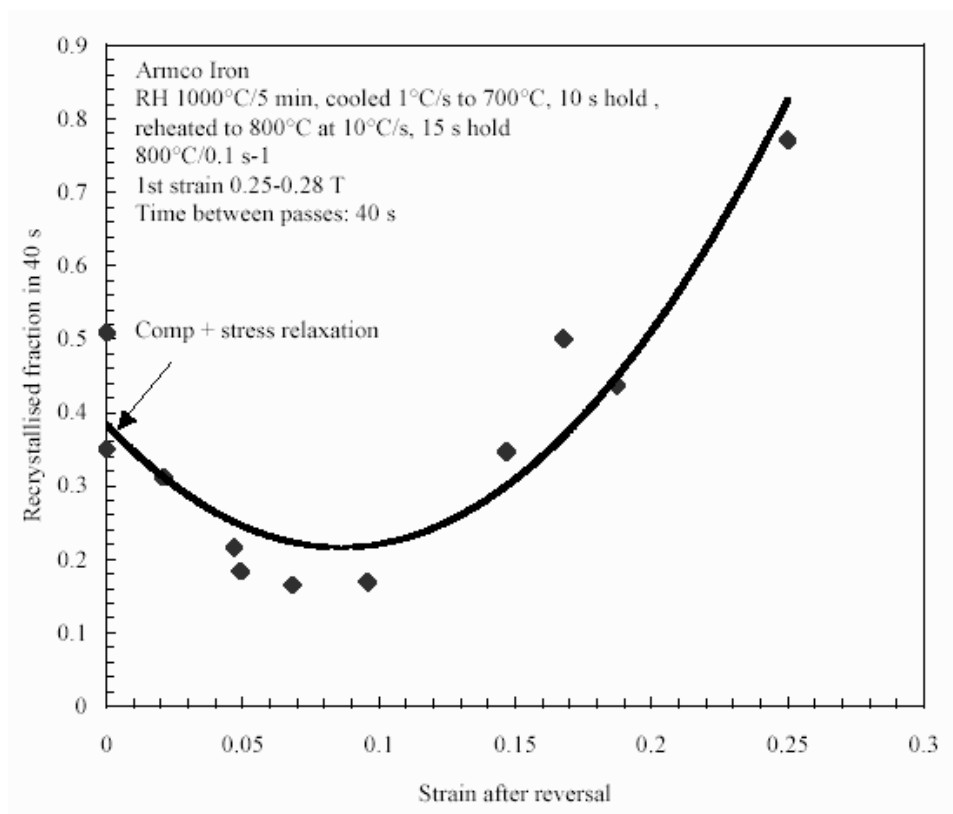
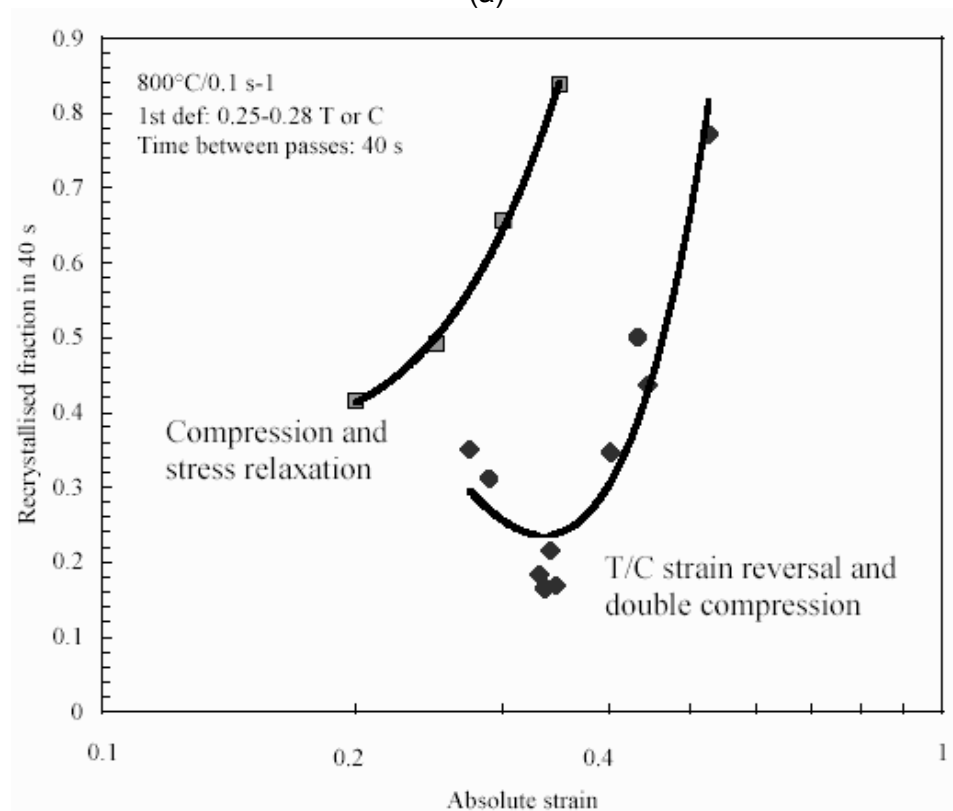


Fig. A6.33: Flow stress in a strain reversal test followed by a second compression pass after 40 s

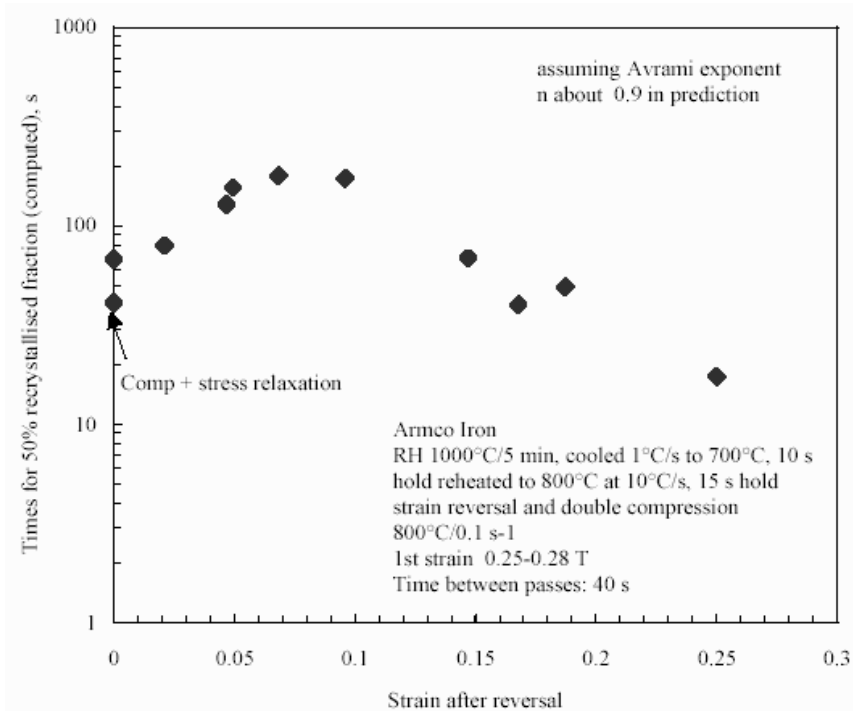


(a)

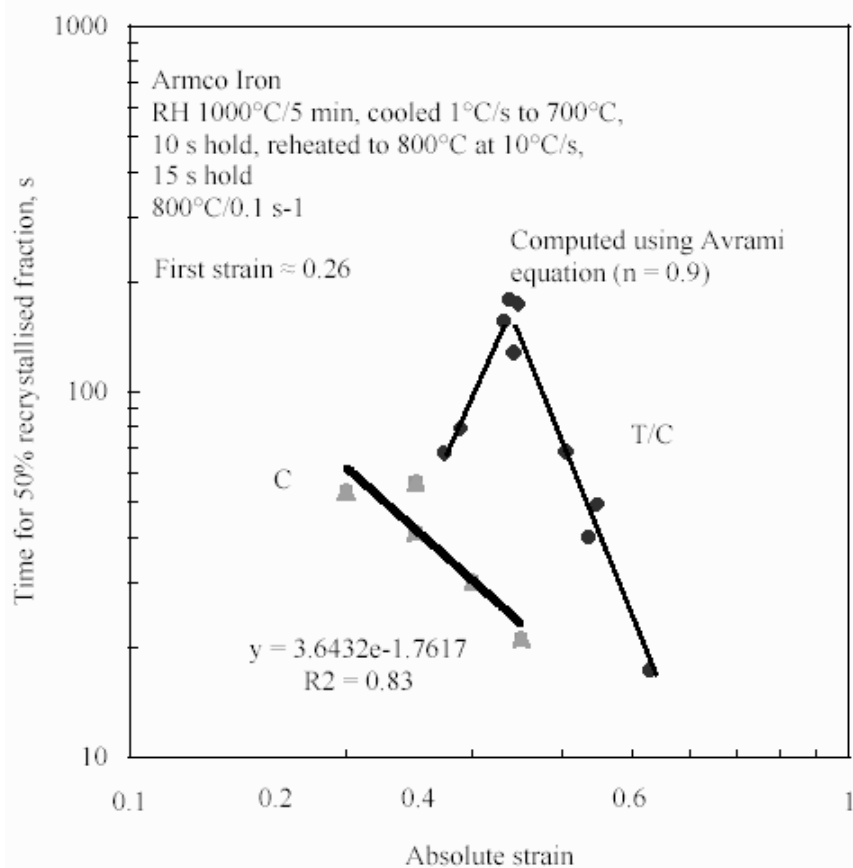


(b)

Fig. A6.34(a and b): Recrystallisation within 40 s as a function of (a) second strain (b) absolute/total strain (also in monotonic compression)

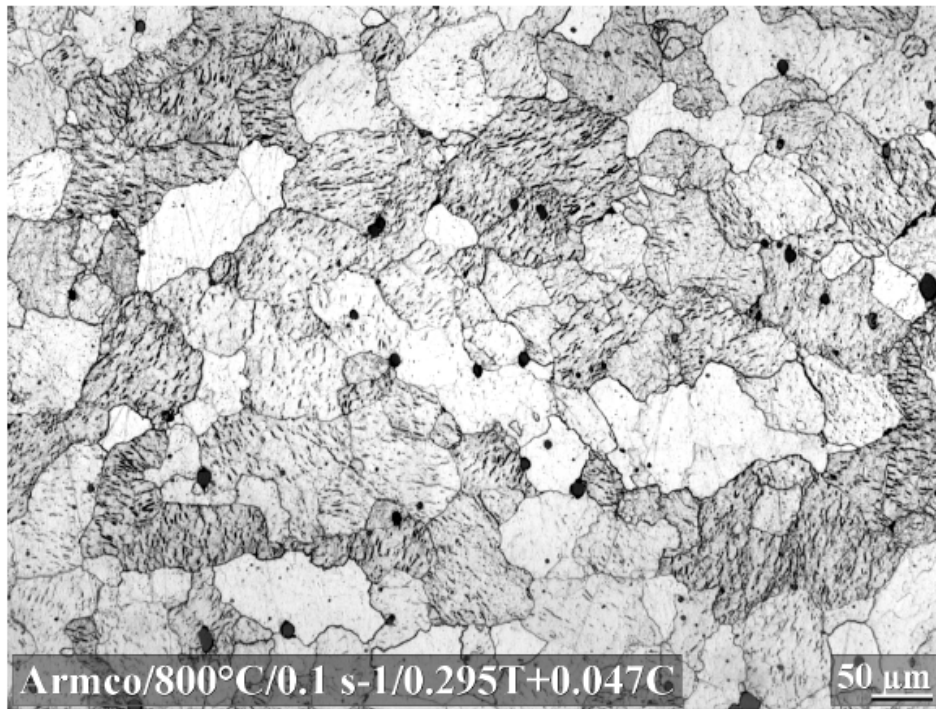


(a)

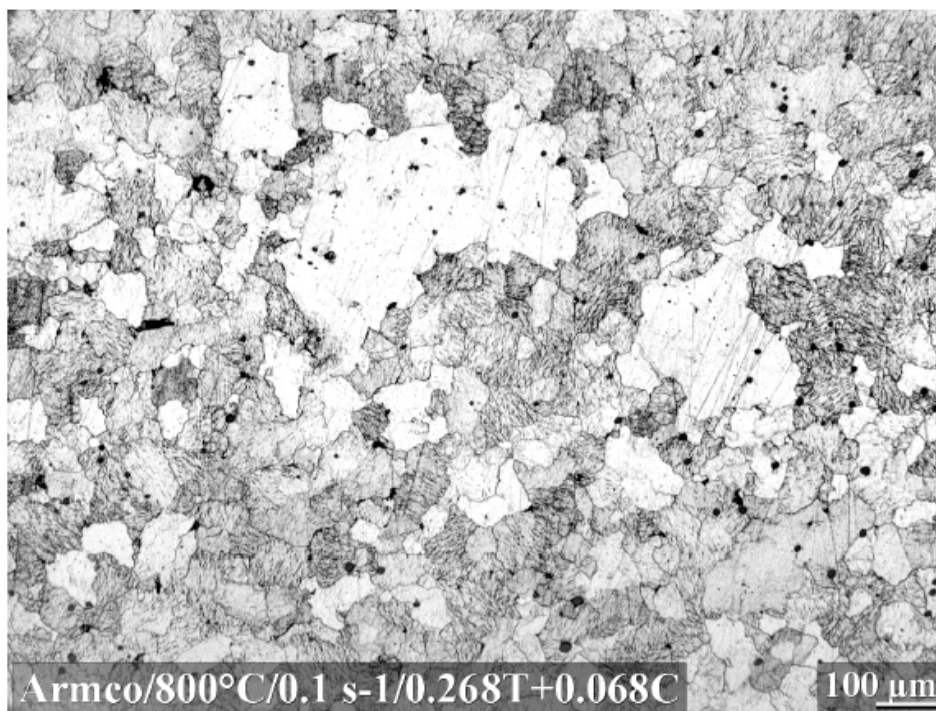


(b)

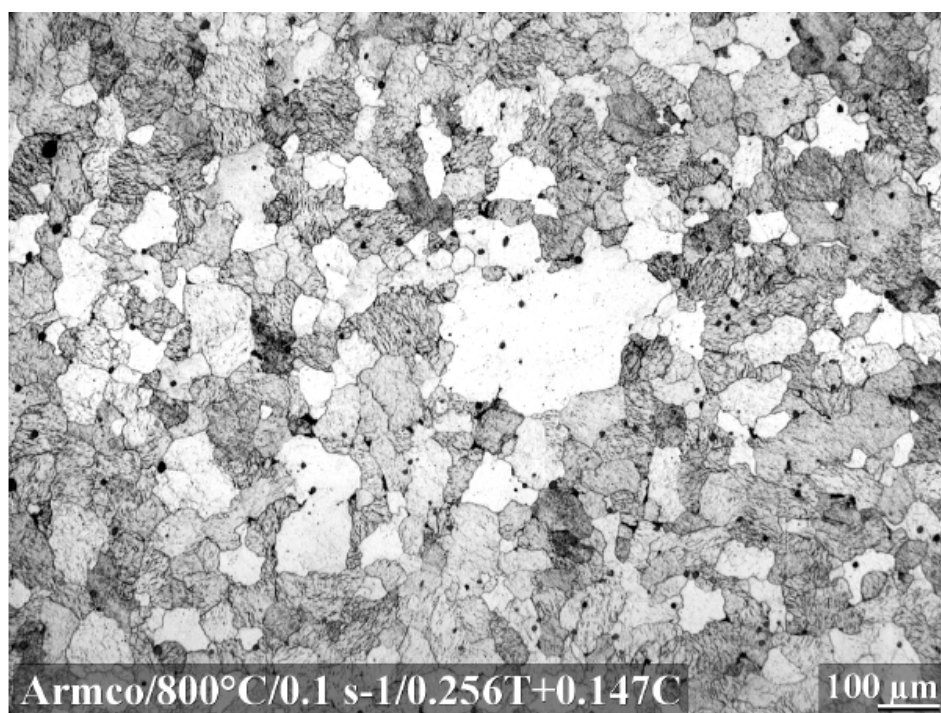
Fig. A6.35(a and b): Predicted t_{50} time as a function of
(a) second strain, (b) absolute/total strain (also in monotonic compression)



(a)

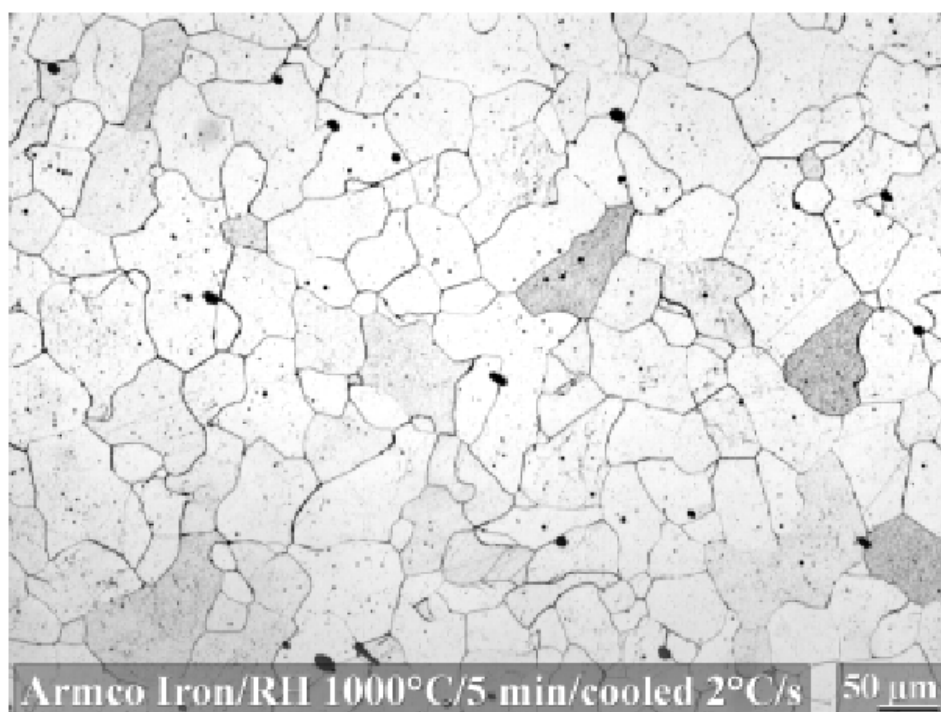


(b)

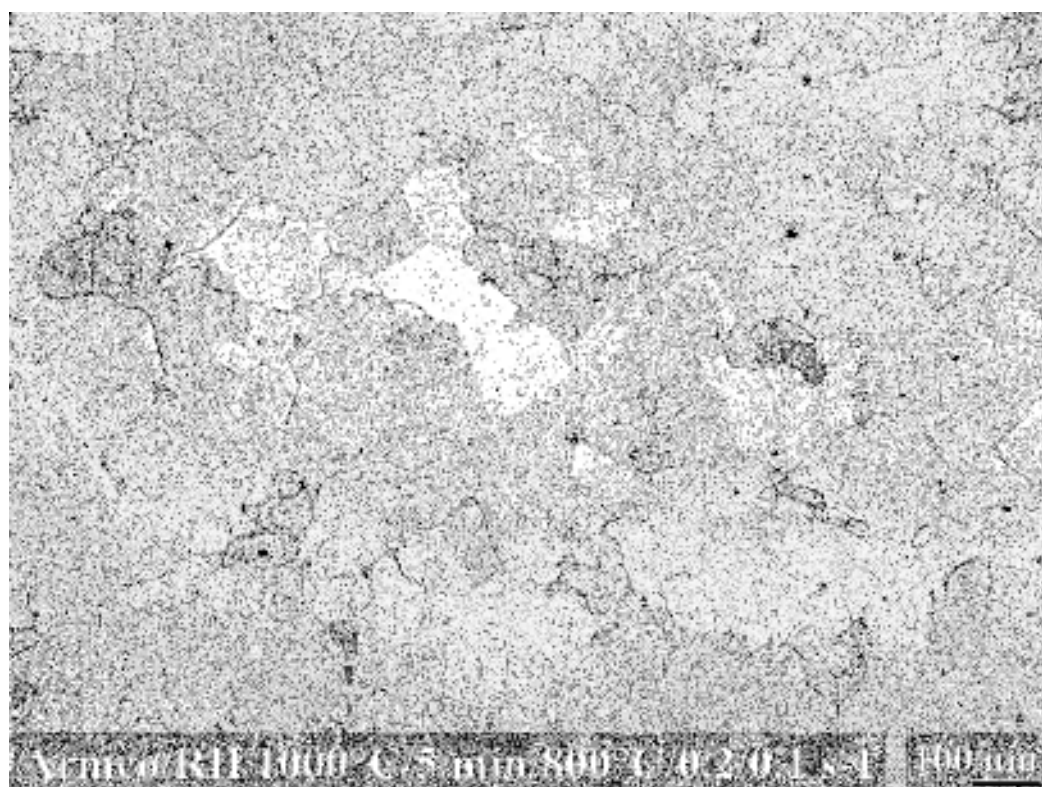


(c)

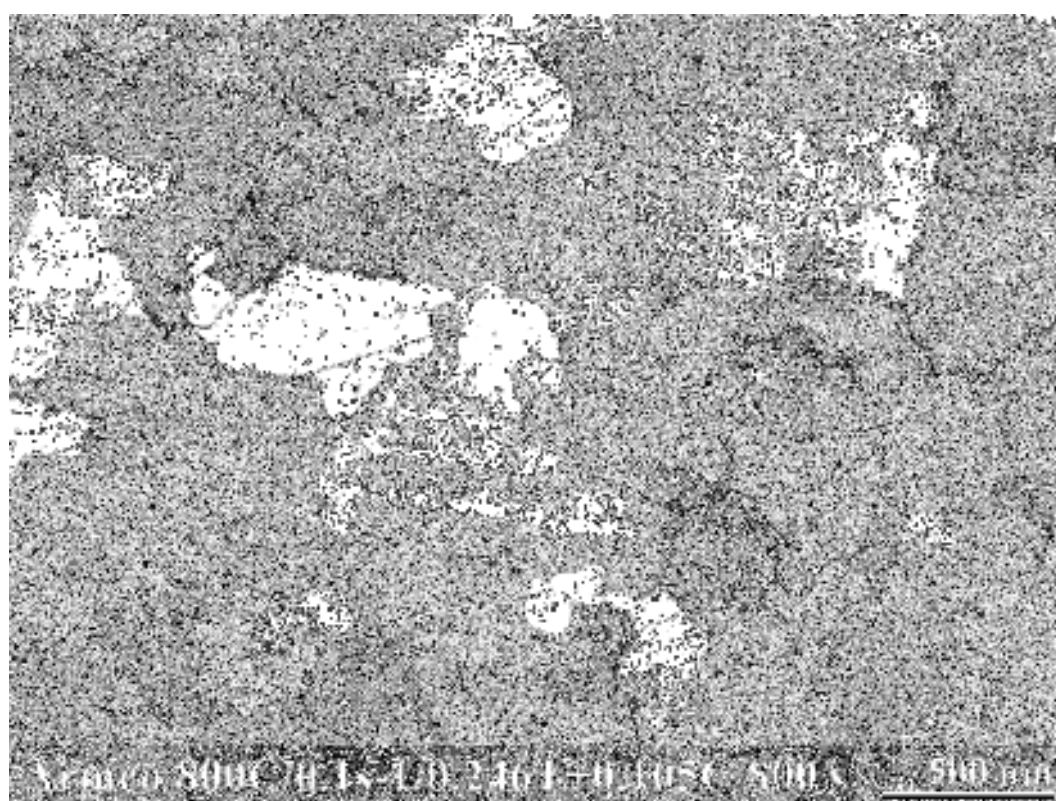
Fig. A6.36(a-c): Grain structure after 40 s revealing large grains after certain reversed strain
 (a) 0.29T+0.05C (b) 0.27t+0.07C (c) 0.26t+0.15C



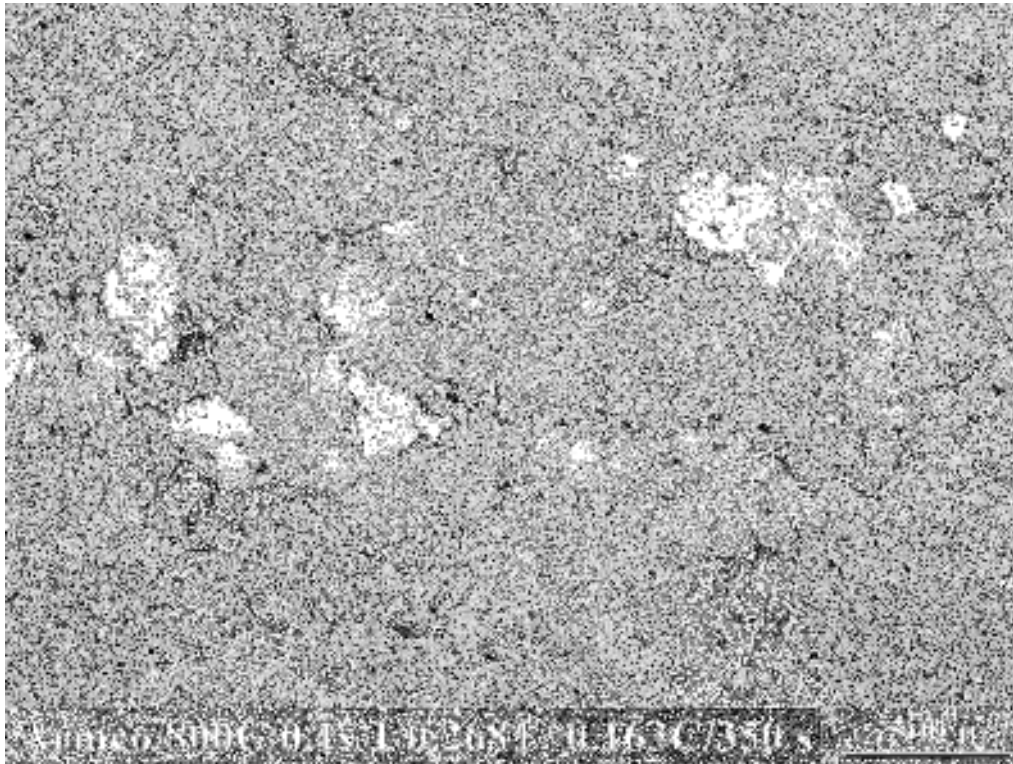
(a)



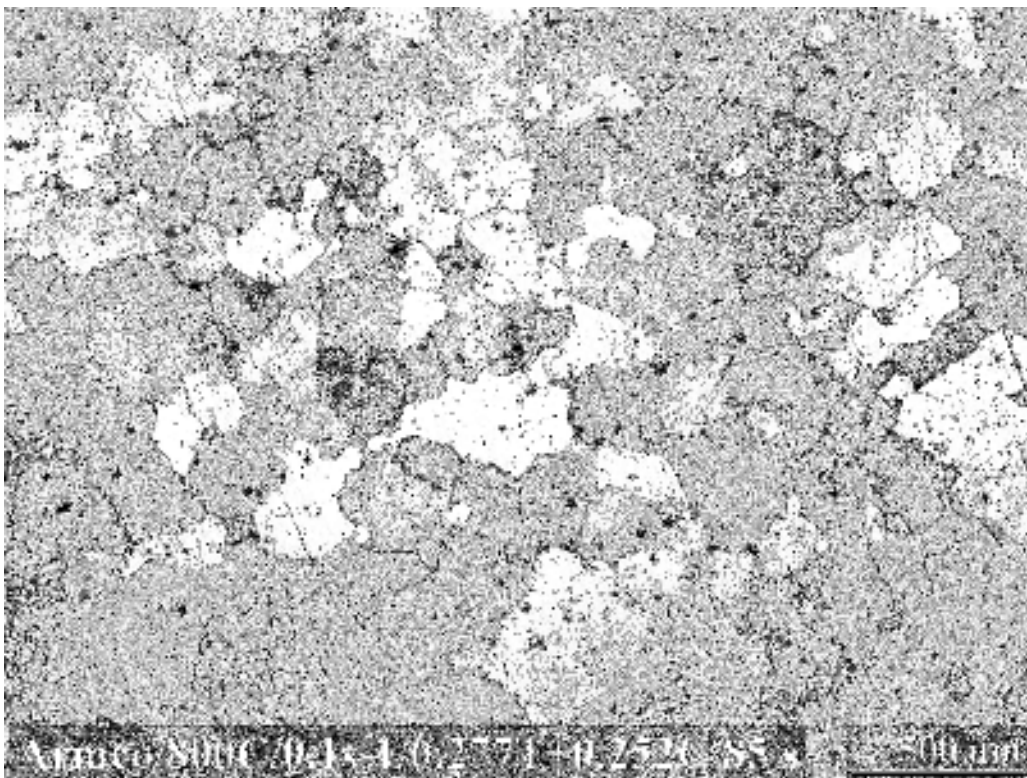
(b)



(c)



(d)



(e)

Fig. A6.37(a-e): Recrystallised grain structure after
(a) without strain (b) 0.20C (c) 0.25t+0.11C (d) 0.27t+0.16C (e) 0.28t+0.25C

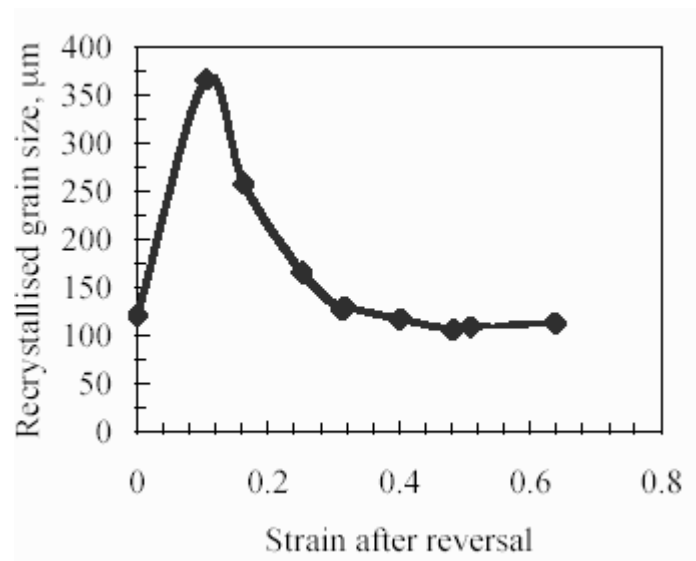


Fig. A6.38: The ferrite grain size as a function of the second strain after the reversal

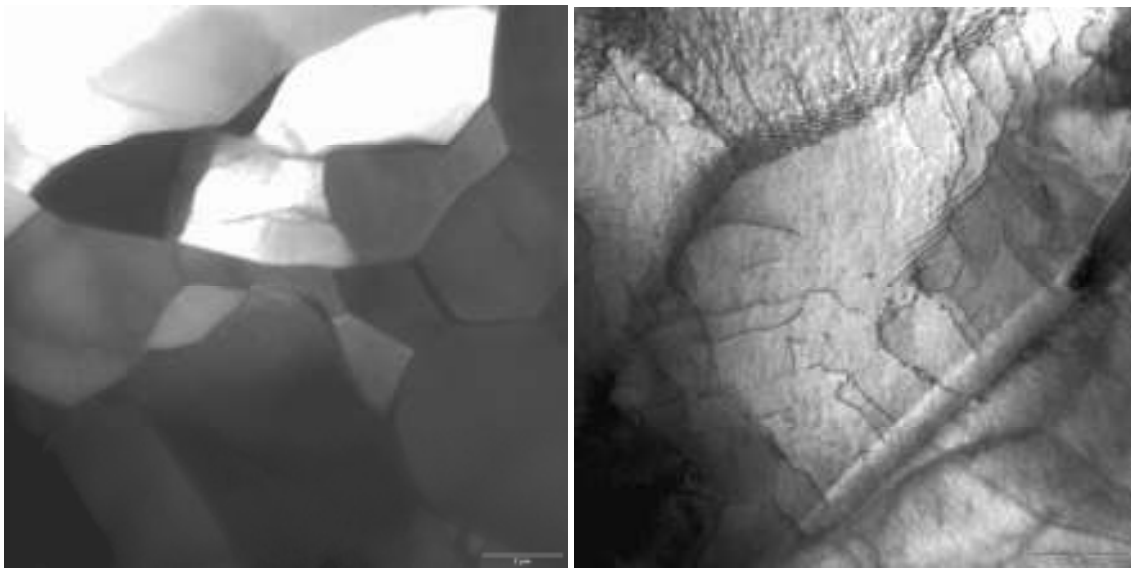


Fig. A6.39: Cell structure in Armco-iron after compression to 0.3 strain

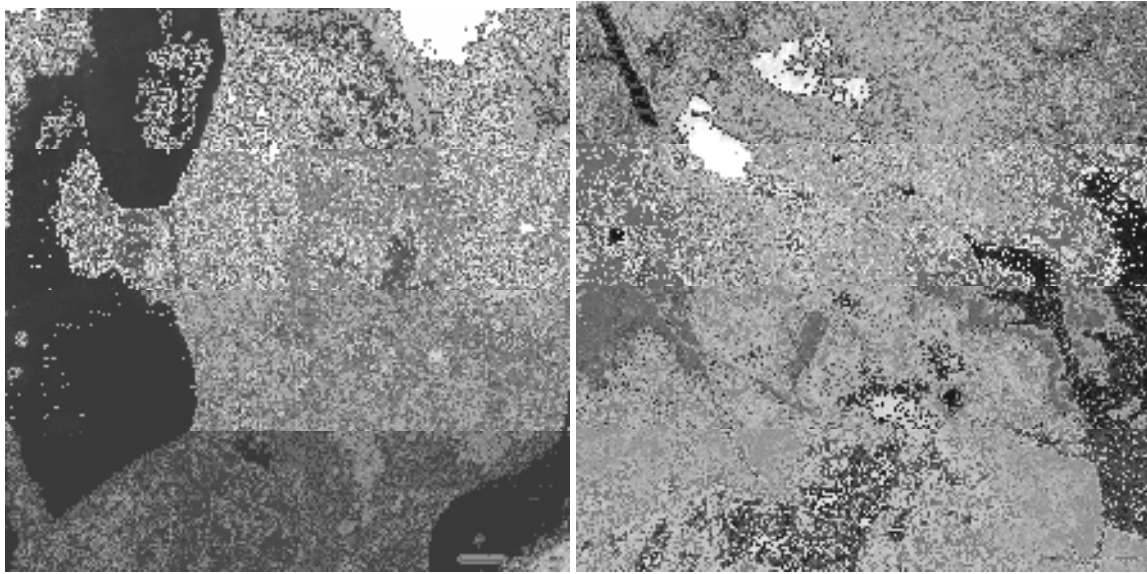


Fig. A6.40: Cell structure in Armco-iron after reversed straining

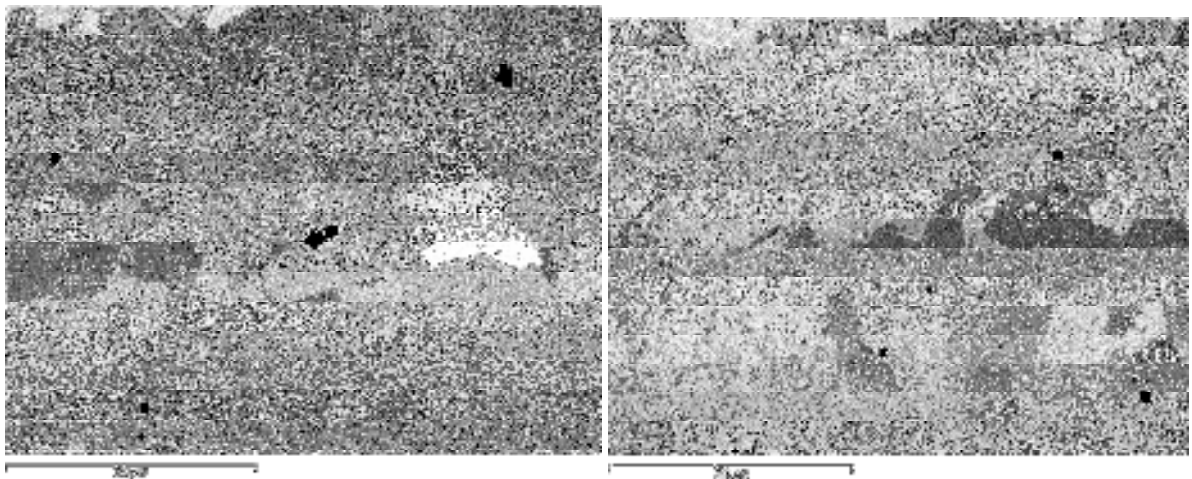


Fig. A6.41: Cell structure in Armco iron seen in SEM after compression to 0.3 strain

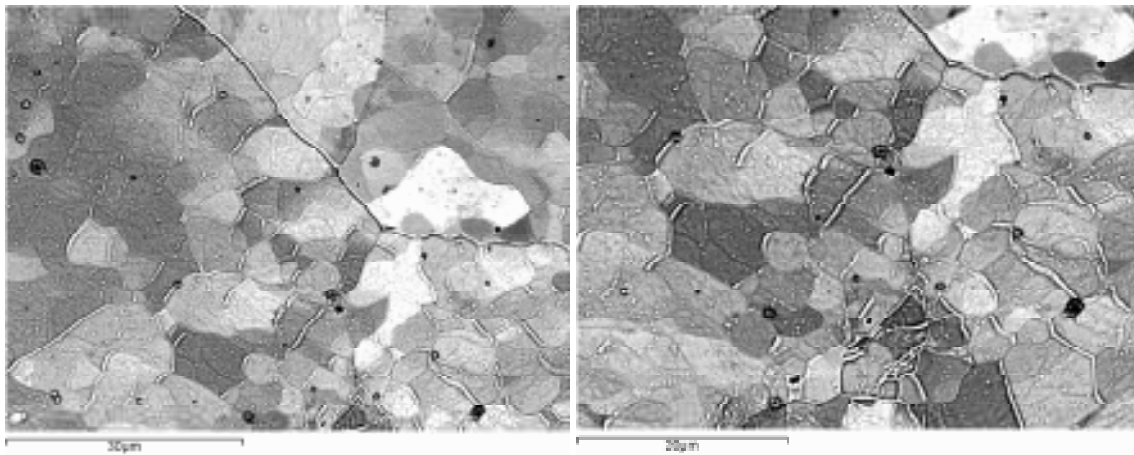


Fig. A6.42: Cell structure in Armco iron seen in SEM after a strain reversal

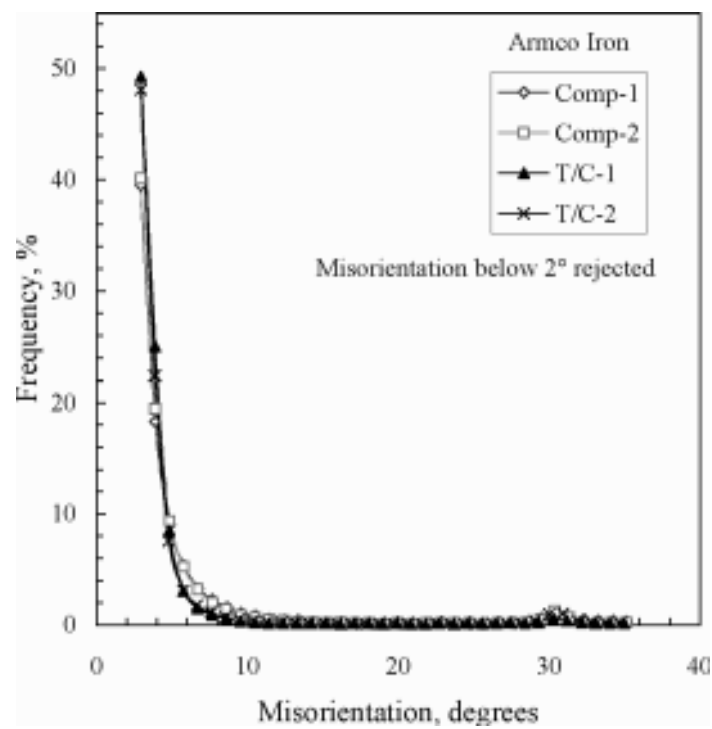


Fig. A6.43: Misorientation distribution with and without strain reversal (comp=compression; t/c=tension/compression; repeated tests for the both).

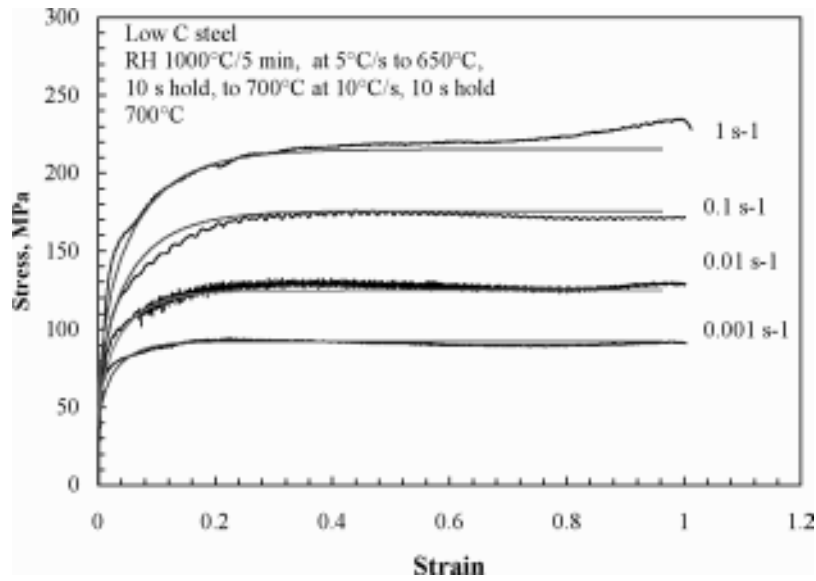


Fig. A6.44: Measured and modelled flow stress curves at constant strain rates

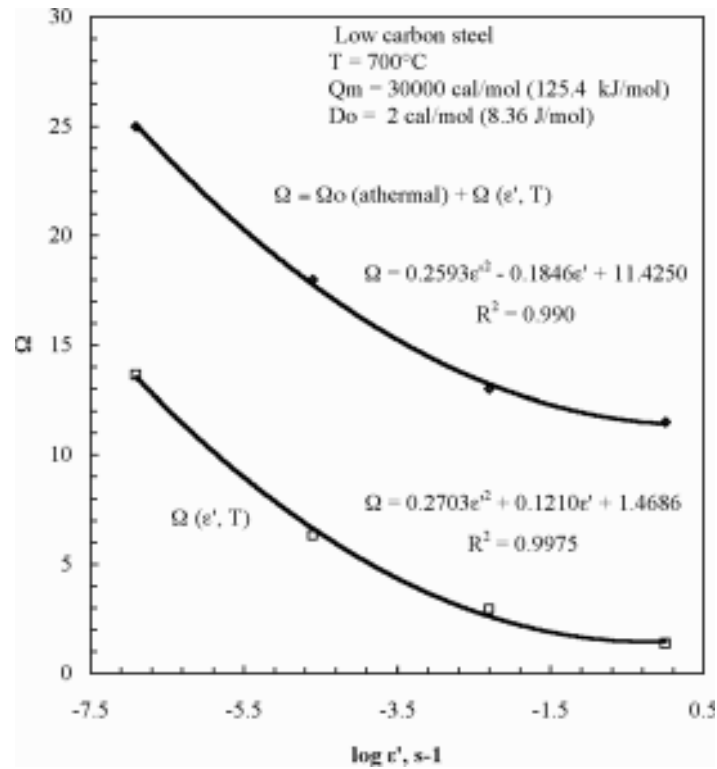


Fig. A6.45: The recovery parameter, the thermal and athermal components as a function of the strain rate

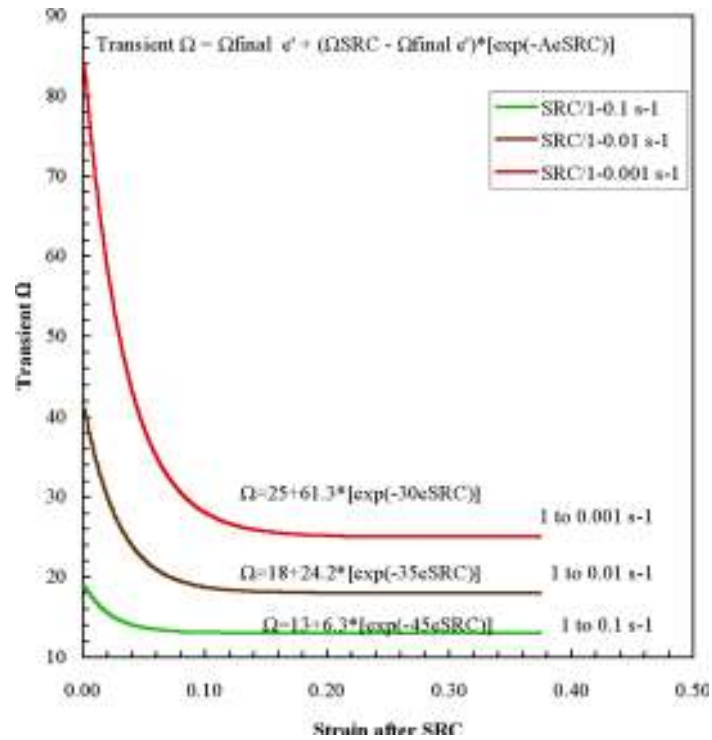


Fig. A6.46: Modelled course of the recovery parameter during the transient

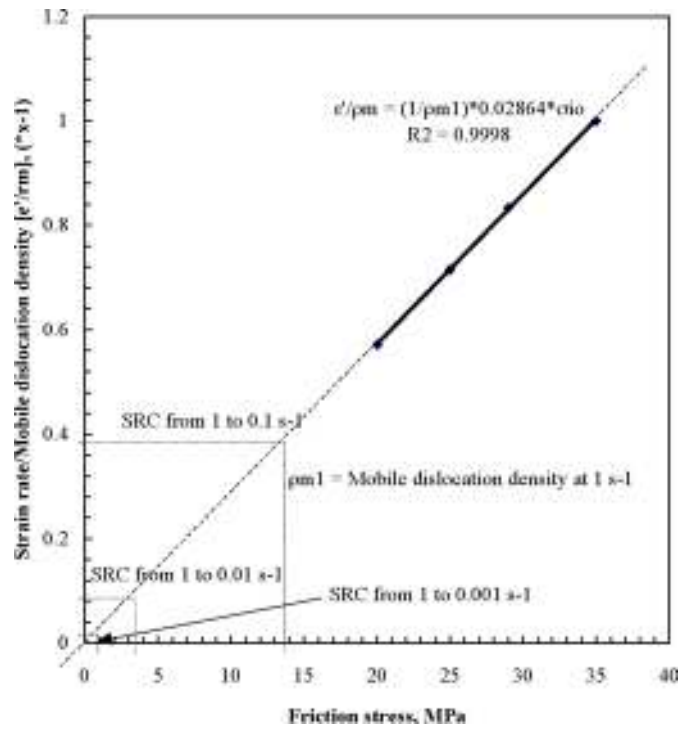


Fig. A6.47: Estimation of the friction stress

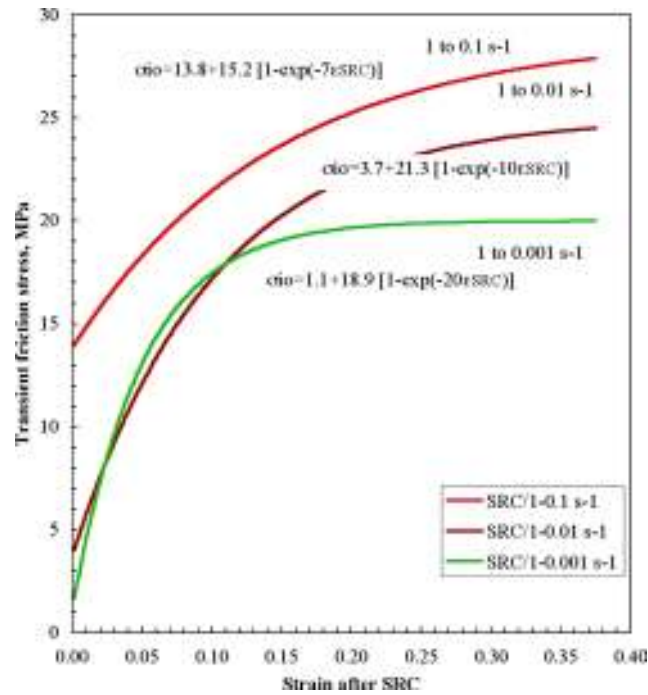


Fig. A6.48: Course of the friction stress during the transient

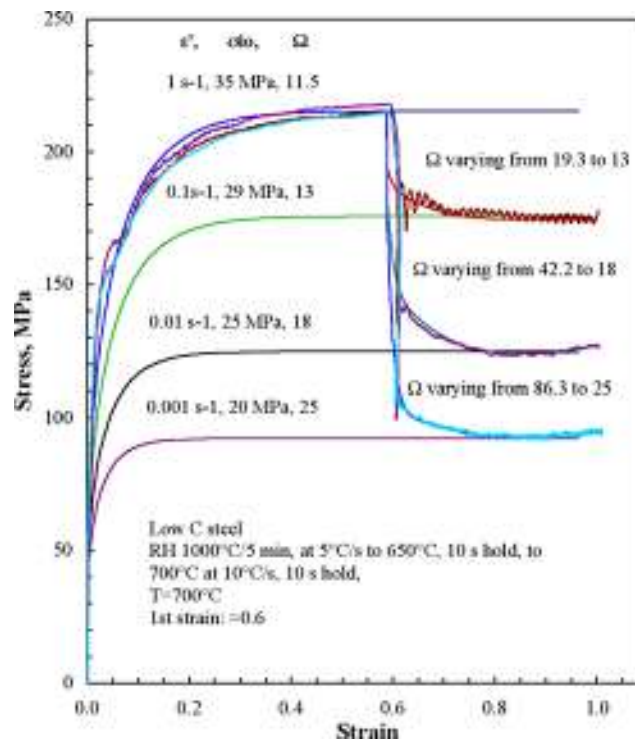


Fig. A6.49: Measured and modelled flow stress curves with the transient (σ_{0i} friction stress, Ω recovery parameter)

Appendix 7

Materials and processing conditions modelled or tested by the various partners

Table A7.1: Materials and processing conditions modelled or tested by the various partners

ECSC P 4405 testing and modelling performed

Material	Armco iron			Low C			Medium C		
	$\Delta T, \Delta \varepsilon'$	FS	SRX/DRX	FS	SRX/DRX	SRV	$\Delta T, \Delta \varepsilon'$	FS	SRX/DRX
complex forming process									
C or C/T									
T (°C) <i>austenite</i>									
strain-range									
str rate (1/s)									
T (°C) <i>ferrite</i>									
strain range									
str rate (1/s)									
metallography									
Torsion									
T (°C) <i>austenite</i>									
strain-range									
str rate (1/s)									
metallography									
T (°C) <i>ferrite</i>									
strain range									
str rate (1/s)									
Rolling									
T (°C) <i>austenite</i>									
strain-range									
str rate (1/s)									
metallography									
Models									
<i>austenite</i>									
<i>ferrite</i>									

DT=temp change, De'=strain rate test, C=compr., C/T=compr/tension, FS=flow stress, SRV=strain reversal, SRX/DRX=static/dynamic recryst.
 OU=Oulu University Fr=Technical University of Freiberg
 * =estimated value# =under development~ = supplier of steel

Material	IF Ti			IF Nb-Ti		
	$\Delta T, \Delta \epsilon'$	FS	SRX/DRX	$\Delta T, \Delta \epsilon'$	FS	SRX/DRX
complex forming						
process						
C or C/T						
T (°C) <i>austenite</i>						
strain-range						
str rate (1/s)						
T (°C) <i>ferrite</i>						
strain range	800			800		
	0.3/0.02			0.29/0.01		
	1/0.01			1/0.01		
str rate (1/s)						
metallography			OU			OU
Torsion						
T (°C) <i>austenite</i>						
strain-range						
str rate (1/s)						
metallography						
T (°C) <i>ferrite</i>						
strain range						
str rate (1/s)						
Rolling						
T (°C) <i>austenite</i>						
strain-range						
str rate (1/s)						
metallography						
Models						
<i>austenite</i>						
<i>ferrite</i>						

DT=temp change, De'=strain rate test, C=compr., C/T=compr/tension, FS=flow stress, SRV=strain reversal, SRX/DRX=static/dynamic recryst.
OU=Oulu University Fr= Technical University of Freiberg
*=estimated value# =under development~ = supplier of steel

Material	C-Mn-S			
	ΔT ,	$\Delta \epsilon$: Slow/Fast Strain Rate		SRV
complex forming	FS	SRX DRX	FS	SRX/DRX
process				
C or C/T				
T (°C) <i>austenite</i>				
strain-range				
str rate (1/s)				
T (°C) <i>ferrite</i>				
strain range				
str rate (1/s)				
metallography	Fr	Fr		
Torsion				
T (°C) <i>austenite</i>				
strain-range				
str rate (1/s)				
metallography				
T (°C) <i>ferrite</i>				
strain range				
str rate (1/s)				
Rolling				
T (°C) <i>austenite</i>	850 1100	850 1100	850 900 1000	850 900 1000 1020 1050
strain-range	0.5	0.5	0.2 - 0.7	0.2 - 0.7 0.15 0.17 0.21
str rate (1/s)	60-140/370-740*	60-140/370-740*	12 - 19	12 - 19 8.7 9.6 10.1
metallography	Fr	Fr	Fr	Fr Corus
Models				
austenite	Fr		Fr	
ferrite				

DT=temp change, De'=strain rate test, C=compr., C/T=compr/tension, FS=flow stress, SRV=strain reversal, SRX/DRX=static/dynamic recryst.
 OU=Oulu University Fr= Technical University of Freiberg
 * =estimated value# =under development~ = supplier of steel

Material	C-Mn-Nb			
	ΔT (Temperature Variation) $\Delta \epsilon$ Slow/Fast Strain Rate	SRV		
complex forming				
process	FS	SRX DRX	FS	SRX/DRX
C or C/T				
T (°C) <i>austenite</i>			1000 1050 1100	1050 1000 1050 1100
strain-range			0.2/0.2 & 0.4C	0.17/0.012-0.8 0.2/0.2
str rate (1/s)				0.1 1
T (°C) <i>ferrite</i>				
strain range				
str rate (1/s)				
metallography	Fr Corus~		Corus	OU Corus~
Torsion				
T (°C) <i>austenite</i>			850-1200	850-1200
strain-range			0.1	0.1
str rate (1/s)			0.01,0.1,1.0	0.01,0.1,1.0
metallography	Fr	Mefos#	CEIT	CEIT
T (°C) <i>ferrite</i>				
strain range				
str rate (1/s)				
Rolling				
T (°C) <i>austenite</i>	850 1100 #	850 1100	850 900 1000 1020 1050	850 900 1000 1020 1050 #
strain-range	0.5	0.5	0.2 - 0.7 0.15 0.17 0.21	0.2 - 0.7 0.15 0.17 0.21
str rate (1/s)	60-140/370-740*	60-140/370-740*	12 - 19 8.7 9.6 10.1	12 - 19 8.7 9.6 10.1
metallography	Fr	Fr	Fr Corus	Fr Corus
Models				
austenite	Fr Mefos#		Fr CEIT	Mefos#
ferrite				

DT=temp change, De'=strain rate test, C=compr., C/T=compr/tension, FS=flow stress, SRV=strain reversal, SRX/DRX=static/dynamic recryst.
OU=Oulu University Fr=Technical University of Freiberg
*=estimated value# =under development~ = supplier of steel

Material	C-Cr-V				C-Mn-Nb-hTi			
	ΔT (Temperature Variation) $\Delta \varepsilon$		SRV		$\Delta T, \Delta \varepsilon'$		SRV	
complex forming	FS	SRX DRX	FS	SRX/DRX	FS	SRX/DRX	FS	SRX/DRX
process								
C or C/T								
T (°C) <i>austenite</i>					1050	1050	1100	1100
strain-range					0.02/0.3/0.02	0.02/0.3/0.02		0.175/0.02-0.4
str rate (1/s)					0.01/1/0.01	0.01/1/0.01		
T (°C) <i>ferrite</i>								
strain range								
str rate (1/s)								
metallography								OU
Torsion								
T (°C) <i>austenite</i>								
strain-range								
str rate (1/s)								
metallography								
T (°C) <i>ferrite</i>								
strain range								
str rate (1/s)								
Rolling								
T (°C) <i>austenite</i>	800 -1200			800 -1200				
strain-range	0.05 - 0.5			0.05 - 0.5				
str rate (1/s)	0.8 3.5 10 30 100			0.8 3.5 10 30 100				
metallography	Mefos			Mefos				
Models								
austenite	Mefos			Mefos#				
ferrite								

DT=temp change, De'=strain rate test, C=compr., C/T=compr/tension, FS=flow stress, SRV=strain reversal, SRX/DRX=static/dynamic recryst.
 OU=Oulu University Fr=Technical University of Freiberg
 *=estimated value# =under development~ = supplier of steel

Material	Type 316 Stainless Steel				Duplex Stainless Steel (As Cast AC, Wrought W.)			
	$\Delta T, \Delta \epsilon'$		$\Delta T, \Delta \epsilon'$		$\Delta T, \Delta \epsilon'$		$\Delta T, \Delta \epsilon'$	
complex forming	FS	SRX/DRX	FS	SRV	FS	SRX/DRX	SRV	SRV
process								Multipass
C or C/T								
T (°C) austenite				1000 1100				
strain-range				0.25 - 0.77				
str rate (1/s)				1				
T (°C) ferrite								
strain range								
str rate (1/s)								
metallography				Corus				
Torsion								
T (°C) austenite						CEIT	CEIT	CEIT
strain-range						950, 1000, 1050 ($\gamma+\delta$ range)	950 1000 1050 ($\gamma+\delta$ range)	1000 ($\gamma+\delta$ range)
str rate (1/s)						0.06 to 0.87	0.06-0.22(AC), 0.87 (W)/0.06-fracture	(0.06/-0.06) or (0.17/-0.17) repeated > 3 times
metallography						0.1, 1.0	0.1 1.0	1
T (°C) ferrite						CEIT	CEIT	CEIT
strain range								
str rate (1/s)								
Rolling								
T (°C) austenite				1100			1150 - 950	
strain-range				0.24 - 0.25			0.18-0.35	
str rate (1/s)				5.4 - 5.9			3 - 10	
metallography				Corus		CSM~		
Models								
austenite			Corus	Corus		CSM	CSM	
ferrite						CSM	CSM	

DT=temp change, De'=strain rate test, C=compr., C/T=compr/tension, FS=flow stress, SRV=strain reversal, SRX/DRX=static/dynamic recryst.
OU=Oulu University of Applied Sciences, Fr=Technical University of Freiberg*=estimated value# =under development~ = supplier of steel

Appendix 8

Technical Annex

1. OBJECTIVES

One of the principal requirements for accurate thermomechanical process modelling is reliable constitutive data for the materials being processed. Such data are normally acquired in laboratory tests at constant strain rate and temperature but with the advent of finite element (FE) modelling techniques it is now possible to adopt more realistic assumptions about the complex loading paths occurring in most metal forming processes. These complex-loading paths may involve non-uniform strain rate, non-uniform temperature, strain reversal and the initiation of defects. The proposed project should be seen in a wider context as contributing to the development of a through-process modelling capability, which will enable optimization of process variables to provide the required microstructural evolution and final product properties.

The objectives of the proposed programme are:

1. To investigate the constitutive behaviour of steels by means of laboratory test simulation of complex loading paths such as strain reversal, variation of strain rate and variation of temperature during deformation.
2. To identify the mechanisms responsible for the constitutive behaviour observed during the laboratory tests by quantifying the sub-structure obtained in terms of dislocation density, cell size, sub-grain boundary misorientation and recrystallisation behaviour using electron optical techniques such as TEM and EBSD where appropriate.
3. To carry out laboratory scale hot rolling trials to investigate the effects on constitutive behaviour of complex strain paths such as those in forward and reverse rolling, horizontal-vertical rolling sequences and very high strain rate processes such as rod rolling.
4. To formulate and further develop alternative constitutive models for predicting flow stress, microstructural evolution, damage initiation and evolution. Also to compare alternative constitutive modelling approaches and identify the advantages and disadvantages of each approach for a particular application.
5. To develop appropriate methodologies and incorporate the constitutive models into finite element models of rolling and to validate the models by comparison with measurements taken in full scale hot rolling mills.

2. WORK PROGRAMME AND DISTRIBUTION OF TASKS

Each sub-section below describes a task which is referred to in the Programme Bar Chart.

2.1 Evaluation of Existing Constitutive Models (All Partners)

It is important to establish a common basis of knowledge between all partners, taking account of different approaches and identifying deficiencies in modelling capability against

future requirements. It is also necessary to identify areas where the different modelling approaches may be compared so that any advantages or limitations may be determined.

It is also proposed to carry out a benchmark simulation exercise in which a mutually agreed metal forming process, e.g. hot rolling of a specified steel grade, is modelled and the results compared to illustrate the relative importance of flow stress measurement, constitutive model and FE model on the accuracy of final predictions of quantities such as rolling load and strain distribution.

2.2 Experimental Investigation of Constitutive Behaviour

2.2.1 Influence of Strain Rate and Temperature Changes (Oulu)

Axisymmetric compression tests will be carried out in a Gleeble Thermomechanical Simulator and in which the strain rate is changed abruptly or continuously or the deformation temperature is changed rapidly. The transient behaviour of the flow stress following the strain rate or temperature changes will be examined and the subsequent static or post dynamic recrystallisation kinetics will be determined. Both the stress relaxation and double compression testing techniques will be employed. Ti-Nb HSLA and medium carbon steels will be investigated.

2.2.2 Influence of Strain Reversal in Tension and Compression (Oulu and Corus UK Limited)

The influence of strain reversal on flow stress and subsequent recrystallisation behaviour will be investigated by means of axisymmetric, alternate tension/compression tests in the Gleeble Thermomechanical Simulators at Corus UK Limited and Oulu University.

The various tensile/compressive sequences will be followed by stress relaxation in order to observe any differences in recrystallisation behaviour. The specimen microstructures will be examined and characterised as described in Section 2.3 below in order to identify the mechanisms responsible for differences in constitutive behaviour. Oulu will investigate Ti-Nb HSLA and medium carbon steels. Corus UK Limited will investigate CMn, CMn-Nb and 316 stainless steels.

2.2.3 Influence of Strain Reversal in Torsion (CEIT)

Torsion tests will be used to provide information about constitutive behaviour following strain reversal. The tests will be carried out for a large set of pre-strain values after which the full flow stress curve for the material can be obtained and the effect of reversal on static and dynamic recrystallisation determined. Hot torsion tests in single pass, multipass and multipass plus strain reversal will be carried out. The materials investigated will be CMn, CMn-Nb and stainless steels.

2.2.4 Bi-Directional Strain Tests (Corus UK Limited)

A novel method of imparting bi-directional strain to a compression sample in the Gleeble Thermomechanical Simulator (strains at 90° to each other) will be investigated as a means of simulating deformation in horizontal-vertical rolling. Deformation sequences will be carried out using uni-directional and bi-directional strains interrupted by quenching to study the microstructure changes during testing. Recrystallisation behaviour will also be determined following the two modes of straining. The specimen microstructures will be

characterised as described in Section 2.3 below in order to rationalise any differences in recrystallisation behaviour. The material investigated will be 316 stainless steel.

2.2.5 Ferritic Tests (Oulu)

In addition to more conventional testing at around 1000-1100°C, a series of tests (both strain reversal and changing strain rate) will be conducted at lower temperatures where the microstructure of the steel is ferritic. The behaviour of ferrite has not been examined earlier in such tests but it is relevant in relation to ferritic rolling for example. The observed phenomena in the ferritic phase will be compared with those in the austenite phase. The materials investigated will be CMn-Nb and medium carbon steels.

2.2.6 Horizontal-Vertical Rolling (Corus UK Limited)

Laboratory rolling trials will be carried out to investigate the differences in recrystallisation behaviour between horizontal-horizontal and horizontal-vertical rolling sequences to the same equivalent strains. In the former pass sequence the strains in successive passes are of the same sign whereas in the latter sequence the strains are reversed in successive passes. The material investigated will be 316 stainless steel.

2.2.7 Forward-Reverse Rolling (TU Freiberg)

The effect of strain reversal will be investigated by forward and reverse rolling. The development of microstructure and mechanical properties will be compared with those obtained by forward rolling only. The effect of forward and reverse rolling on the development of thermal distributions, which may affect the microstructural evolution, will be investigated for selected materials.

2.2.8 High Speed Rolling (TU Freiberg and Corus UK Limited)

The experimental rod mill at TU Freiberg provides the opportunity to derive information about the constitutive behaviour of steels at very high strain rates, i.e. of the order of 1000 s⁻¹. Such strain rates are not achievable with conventional laboratory test equipment. However, it is possible, that information about the constitutive behaviour of steels at such high strain rates may be inferred from experimental rolling data by use of a suitable theoretical process model. This approach is to be investigated.

2.3 Characterisation of Structures and Metallographic Analysis (Corus UK Limited, Oulu and CEIT)

In the laboratory tests described above it is important that a detailed understanding of the fundamental mechanisms responsible for the material constitutive behaviour is developed. This is necessary in order to formulate physically meaningful constitutive models. Therefore, characterization of the microstructures will be accomplished using electron optical methods (TEM/EBSD) where appropriate, i.e. for stainless steels or deformed ferritic structures, to quantify the sub-structure in terms of dislocation density, cell size and sub-grain boundary misorientation.

2.4 Formulation of Constitutive Laws (All Partners)

All partners will be involved with the formulation of constitutive laws using a number of alternative approaches.

For example MEFOS will develop further the dislocation density based model for predicting flow stress under varying strain rate conditions and taking account of dynamic recrystallisation. Corus UK Limited will investigate the physical-phenomenological modelling approach. CSM will develop constitutive equations for multiaxial loading, including damage evolution, for use in the three dimensional case of long product rolling. CEIT will develop a physical/semi-empirically based constitutive model to describe the effect of strain reversal on the flow stress curves and on the static and dynamic recrystallisation. TU Freiberg will make further developments to existing constitutive models, e.g. Spittel-ansatz and dislocation density models.

2.5 Incorporation of Constitutive Laws into FE Models of Rolling (Corus UK Limited, CSM, MEFOS and TU Freiberg)

The constitutive models formulated in Section 2.4 above will be incorporated into FE models of rolling.

Emphasis will be placed on developing the methodologies for incorporating complex constitutive models into FE models. This is usually accomplished by means of a user defined material subroutine (UMAT).

2.6 Validation of Models in Pilot and Full Scale Rolling Mills (Corus UK Limited, CSM, MEFOS and TU Freiberg)

TU Freiberg, MEFOS and Corus UK Limited have excellent laboratory/pilot scale rolling facilities which are fully instrumented and will provide initial data for validation of the models. TU Freiberg will utilise optical visioplasticity techniques and the analysis of grey scale images will be further developed for measurements of local surface deformations in the roll gap. These measurements will be compared with FE model predictions. Corus UK Limited and CSM have access to full scale industrial hot rolling mills and will provide data for final validation of the models over a wide range of steel compositions and rolling conditions.

3. WAYS AND MEANS

In order to explain the constitutive behaviour observed when the materials are subjected to complex loading paths and in order to facilitate the development of constitutive models, electron optical metallographic techniques such as TEM and EBSD will be employed to characterise the sub-structures obtained in terms of dislocation density, cell size and grain boundary misorientation.

Particular attention will be given to the development of methodologies for incorporating constitutive models into FE models. It is important to ensure that the increased complexity of the models does not make their use impracticable because of the high computing demand.

The exchange of test data and models between partners will enable alternative approaches and procedures to be compared and assessed.

Corus UK Limited will be responsible for the overall management of the project and will appoint a Project Co-ordinator. Each partner will appoint a Technical Manager to liaise with the Project Co-ordinator.

4. PROGRAMME BAR CHART (See next page)

5. The results of the project will be the subject of a publication in the "Technical Steel Research" series.

The research described above will be placed in the area covered by the Executive Committee (or the Expert Group): D3

Task	Work Programme	Partner(s)	Year 1 (1/7/01-30/6/02)		Year 2 (1/7/02-30/6/03)		Year 3 (1/7/03-30/6/04)		Year 4 (1/7/04-31/12/04)
			Semester 1	Semester 2	Semester 3	Semester 4	Semester 5	Semester 6	
2.1	Evaluation of Existing Constitutive Models	All Partners							
2.2.1	Influence of Strain Rate and Temperature Changes	Oulu							
2.2.2	Influence of Strain Reversal in Tension and Compression	Oulu & Corus							
2.2.3	Influence of Strain Reversal in Torsion	CEIT							
2.2.4	Bi-Directional Strain Tests	Corus							
2.2.5	Ferritic Tests	Oulu							
2.2.6	Horizontal-Vertical Rolling	Corus							
2.2.7	Forward-Reverse Rolling	TU Freiberg							
2.2.8	High Speed Rolling	TU Freiberg & Corus							
2.3	Characterisation of Structures and Metallographic Analysis	Corus, Oulu & CEIT							
2.4	Formulation of Constitutive Laws	All Partners							
2.5	Incorporation of Constitutive Laws into FE Models of Rolling	Corus, CSM, MEFOS & TU							
2.6	Validation of Models in Pilot and Full Scale Rolling Mills	Corus, CSM, MEFOS & TU							
General / Reporting									
	Semester Reports (R)		M	M	R	R	R	R	R
	Co-ordination Meetings & Travel (M)		M	M	M	M	M	M	M
	Final Report								

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