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Testing and evaluation of material data for analysis of forming and hardening of boron steel components

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Abstract

Finite element modelling and simulation is becoming an increasingly important tool in the development process for structural automotive components, manufactured using thermo-mechanical forming techniques. Accurate and reliable analysis of coupled thermo-mechanical processes requires efficient simulation tools as well as good quality and relevant material data, usually obtained by experimental testing of the mechanical and thermal properties. The work present in this paper concerns methods for obtaining and evaluating the mechanical properties, required for modelling the high-temperature forming of a high-strength boron-alloyed steel. The material data was obtained from high temperature compression tests and dilatometric measurements made using a Gleeble 1500 thermo-mechanical simulator. Two examples of finite element simulations using the data obtained are also presented. The first example is an isothermal finite element simulation of a thin-walled tubular beam subjected to high-temperature bending. The predicted press force showed acceptable agreement with experimental results in the initial part of the process. In the second example, a cylindrical specimen compressed during continuous cooling was simulated, and the press force and radial displacement were compared with experimental results. Again the simulations showed acceptable agreement with experimental results but indicated the need for further improvements in the simulation technology and methods used for material parameter evaluation.

1. Introduction

Simultaneous forming and quenching of boron-alloyed steel is a process used for the manufacturing of high-strength steel components, such as bumpers and side-impact protection

beams. In the manufacturing process considered, thin-walled profiles of a hardenable steel are formed by shape rolling and subsequently seam welded into closed-section profiles, which are heated to austenite region, placed in a forming tool and formed to the desired shape. While still clamped in the forming tool, these components are subsequently hardened by spray quenching before being released. The physical properties and hence performance of these components, are greatly influenced by the microstructure, residual stresses and geometrical defects imposed by the manufacturing process. To enable efficient development of these products it is therefore necessary to be able to quickly, and reliably simulate and analyse the manufacturing processes using numerical computational tools, e.g. finite element (FE) methods. However, FE simulation of coupled thermo-mechanical processes requires not only efficient computational tools but also accurate material data. Unfortunately, material properties are rarely known for the specific steels used at the temperature and time range required for the martensite transformation characteristic of this hardening process. For this reason, assumptions and simplifications are generally used in the FE models, which may lead to inaccuracies. This paper describes experimental methods used to obtain the mechanical material properties at elevated temperatures for a boron-alloyed steel Docol BO 02, used in a specific forming and hardening process.

2. Simulation model requirements

The manufacturing process of interest in this paper is a thermo-mechanical forming operation. To allow FE modelling of this process, a material model for the steel that includes plastic deformation with a non-linear, temperature-dependent, hardening function is required. The yield stress and elastic modulus, which are also temperature dependent, must also be modelled as non-linear functions over the range of temperatures encountered in the process. The material model should also describe the volume change with temperature as well as the transformation plasticity during the hardening process.

Before forming, the material is heated to a temperature in the austenite region; slightly above 850°C. During and after the forming process the material must be cooled to room temperature quickly enough to ensure that only the martensitic transformation takes place. Time-dependent diffusional effects and grain growth will influence the mechanical properties of the material. Although material models for microstructural predictions including arbitrary thermal histories has been developed [1], in this paper the phase transformation has been modelled as being time independent. For this reason, it is important that deformation and temperature histories comparable to the process being simulated are used when obtaining materials data.

The material description concerning thermal behaviour includes non-linear functions for thermal conductivity and specific heat. The intended material model also takes account of latent heat generation during phase transformation from austenite to martensite. The process model must also include a description of the temperature distribution in the thickness direction of the material and consider the mechanical and thermal contact between the forming tools and the component being formed. The evolving change in the contact geometry as the component is formed is the main reason why a coupled thermo-mechanical analysis of the process is necessary.

Traditional analyses of metal forming and quenching processes, using the finite element method, are covered extensively in the literature (see [2, 3]). However, there are few reports concerning the modelling and simulation of the combined forming and quenching of thin-walled steel components. Early forming simulations using explicit integration of the equations of motion are described in [4]. Comparisons with traditional implicit methods have been made

by several authors (see e.g. [5]). Examples of coupled thermo-mechanical analyses for different types of bulk forming operations can be found in [6–8]. Non-isothermal analyses of sheet metal forming where the temperature rise due to the plastic work is accounted for are described by several researchers (see e.g. [9, 10]). This paper is based on the developments reported in [11], where analysis methods and material modelling have been developed to enable simulation of the combined forming and quenching of steel sheets. The material model that will be used for the coupled thermo-elasto-plastic analyses of the forming and quenching of the tubular beams is described in [12]. This mechanical model is intended to be used together with a non-linear thermal model (see [12]). The volume change during the martensitic transformation is given by the thermal expansion function:

$$\varepsilon^{\text{th}} = \varepsilon^{\text{th}}(T), \quad (2.1)$$

which is defined for the material when subjected to an appropriate cooling history. This function is obtained from the measured dilation curve, which includes both thermal expansion and volume change due to phase transformation during cooling. This function is implemented in the material model by assuming the total strain increment $\Delta\varepsilon^{\text{Tot}}$ to be composed of elastic $\Delta\varepsilon^{\text{e}}$, plastic $\Delta\varepsilon^{\text{p}}$, thermal $\Delta\varepsilon^{\text{th}}$, and transformation plasticity $\Delta\varepsilon^{\text{trp}}$ strain increment such that

$$\Delta\varepsilon^{\text{Tot}} = \Delta\varepsilon^{\text{e}} + \Delta\varepsilon^{\text{p}} + \Delta\varepsilon^{\text{th}} + \Delta\varepsilon^{\text{trp}}. \quad (2.2)$$

The present material model is based on the yield condition according to von Mises yield theory and can be written as

$$f(\bar{\sigma}) = \bar{\sigma} - \sigma_y = 0, \quad (2.3)$$

where $\bar{\sigma}$ is the effective stress given by

$$\bar{\sigma} = \sqrt{3J_2} \quad (2.4)$$

and J_2 is the second deviatoric stress invariant. The corresponding hardening law is defined as

$$\sigma_y = \sigma_y(\bar{\varepsilon}^{\text{p}}). \quad (2.5)$$

The associated von Mises flow rule is described by

$$\dot{\varepsilon} = \dot{\lambda} \frac{\partial f}{\partial \boldsymbol{\sigma}}, \quad (2.6)$$

where $\dot{\lambda}$ is the plastic multiplier, $\dot{\varepsilon}$ is the strain rate vector and $\boldsymbol{\sigma}$ is the stress vector. Equations (2.3), (2.6) and the consistency condition

$$\dot{\lambda} \dot{f} = 0 \quad (2.7)$$

then lead to the elastic plastic stress update relation given by

$$\dot{\boldsymbol{\sigma}} = \left(\mathbf{C} - \frac{\mathbf{C} \mathbf{a}^{\text{T}} \mathbf{a} \mathbf{C}}{(\mathbf{a}^{\text{T}} \mathbf{C} \mathbf{a}) + H'} \right) \dot{\varepsilon}, \quad (2.8)$$

where $\mathbf{C}$ is the standard tangential elastic matrix and $\mathbf{a}$ is the flow vector according to

$$\mathbf{a} = \frac{\partial f}{\partial \boldsymbol{\sigma}}. \quad (2.9)$$

The hardening parameter H' is then defined as

$$H' = \frac{\partial \sigma_y}{\partial \bar{\varepsilon}^{\text{p}}}. \quad (2.10)$$

The numerical implementation is based on the ‘effective-stress’ function according to [13]. The present implementation is described in [14] and incorporates a temperature-dependent

piecewise linear isotropic hardening rule and modelling of transformation plasticity based on [14] and [15]. The hardening law is formulated in incremental form as

$$\sigma_y^{t+\Delta t} = \sigma_y^t + H'^{t+\Delta t} \Delta \bar{\epsilon}^p, \quad (2.11)$$

where t is present time and Δt is the incremental time step size controlled by the time integration routine. Here the hardening parameter H' is described by a piecewise linear hardening law defined by tabulated values of yield stress versus effective plastic strain. These values are extracted directly from the experimental flow stress curves. The aim of this paper was to measure and evaluate the mechanical properties of the material at different temperatures after austenitization, i.e. when the material had been heated above the austenitization temperature, cooled to a specific temperature and then subjected to isothermal deformation.

2.1. Material properties

The required material properties to be determined for inclusion in the material model are listed below.

Strain hardening functions at different temperatures. The functions are defined from the flow stress curves measured at discrete temperatures. In the material model, function values for intermediate temperatures are obtained by interpolation. In this study, the functions and the influence of strain rate have been evaluated from compression tests.

Thermal dilatation due to thermal expansion and phase transformations. The thermal expansion and volume change due to the transformation of austenite into martensite have been evaluated using conventional dilatation tests. The specimen was subjected to forced cooling and the dilatation continuously measured.

Young's modulus as a function of temperature. In order to evaluate strain hardening from the compression tests, Young's modulus at the actual temperature is needed. However, Young's modulus cannot be readily determined from compression tests due to elastic strain in the testing machine and the specimen/anvil structure resulting in poor accuracy in the elastic range. The Young's modulus at different temperatures has thus been taken from published data [11, 16].

Poisson's ratio as a function of temperature. Poisson's ratios for two discrete temperatures were used in the analyses presented in this paper and were obtained from the literature [16].

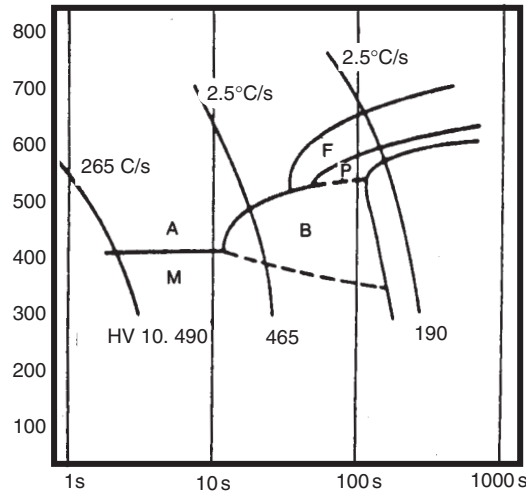
3. Material description

The material studied, Docol BO 02, is a boron steel manufactured by SSAB, Sweden. The chemical composition of this material supplied by the manufacturer is presented in Table 1. For comparison, one specimen was analysed at Rautaruukki Steel, Finland. This composition can also be seen in Table 1. These results indicate that the composition of the material tested is quite typical for the grade, with all elements near the median of the specification.

The continuous cooling transformation (CCT), which indicates the hardenability of the steel, provided by the steel manufacturer is shown in figure 1. From this, it can be seen that a cooling rate greater than about 30°C s^{-1} will result in a martensitic microstructure. According to the technical data, in the hardened state the material has high yield strength (1200 MPa) with an elongation of around 10% at room temperature. At lower cooling rates, bainite or even ferrite can form resulting in lower hardness and strength.

Table 1. Chemical composition of the SSAB Docol BO 02 boron steel.

	C%	Si%	Mn%	P%	S%	Cr%	B%	Al%
Specification	0.20–0.25	0.20–0.35	1.00–1.30	0.030	0.010	0.150–0.250	0.0015–0.005	0
Measured	0.223	0.29	1.11	0.013	0.003	0.211	0.0034	0.051

**Figure 1.** CCT for Docol BO 02 as given in the manufacturer's technical specification.

4. Experimental setup and material properties

4.1. Isothermal compression tests at different temperature ranges

In the manufacturing process of interest, the material is subjected to a temperature history of heating and subsequent cooling at a sufficiently high rate to ensure the formation of martensite. Since the complete thermo-mechanical history of the material will influence its microstructure and properties, appropriate deformation and temperature histories must be used when carrying out the material testing. A Gleeble 1500 thermo-mechanical simulator was employed to create the thermo-mechanical schedules required. The experimental procedure followed was insertion of the cylindrical test specimen in a vacuum chamber, resistance heating to the austenite temperature and subsequent compression between tungsten carbide anvils before controlled cooling. Tantalum foils were used to prevent the specimen sticking to the anvils, and graphite foils for lubrication. The material temperature was measured using a Pt/PtRh thermocouple welded to the surface of the specimen. The axial and radial displacement and the axial compressive force were measured continuously during the experiment. The stress and strain curves presented have been corrected to account for the tantalum and graphite foils and the elasticity of the machine. Different testing techniques had to be developed for each of the different temperature regimes.

As can be seen in figure 1, at high temperatures ($T > 650^{\circ}\text{C}$), the austenite phase is stable for a long time so that no particular requirements exist regarding the cooling history of the specimen. Conventional cylindrical specimens 10 mm dia. $\times$ 12 mm long could therefore be used. The specimens were heated to 900°C , and held at that temperature for 5 min. They were then cooled at a rate of about $10^{\circ}\text{C s}^{-1}$ to the testing temperature, which was held for 15 s.

The specimens were then compressed to a final strain of 0.8 at a true strain rate of 0.1 s^{-1} . To investigate the influence of strain rate, additional tests were performed with higher and lower strain rates.

As seen in figure 1, in the intermediate temperature regime the cooling rate and holding time can have a significant effect on the microstructure of the steel and consequently on its stress-strain behaviour. After several trials with different specimen geometries, compression testing using small cylinders of 4 mm dia. $\times$ 5.5 mm long was used. In these tests, the cooling rate of 50°C s^{-1} could be achieved down to the martensite-start-temperature (M_s 420°C), although the cooling rate decreases during martensite formation. Determination of stress-strain curves for austenite at temperatures above M_s and austenite + martensite above the martensite-finish-temperature ($M_f < T < M_s$), were enabled by immediate compression, using higher strain rates and no holding time.

At lower temperatures, below the martensite-finish-temperature ($M_f < 300^\circ\text{C}$), the microstructure of the steel becomes fully martensitic if the cooling rate has been high enough. Dilatation testing indicated even a shorter time before the start of bainite transformation than shown in figure 1. The required cooling rate is, however, not attained on a Gleeble if using specimens with the dimensions as mentioned above. Therefore, cylindrical rods 5.2 mm in diameter were heated in a furnace to 900°C and water quenched to obtain the correct, i.e. martensitic, structure. The rods were skinned to 5 mm dia. and cut into pieces of 7 mm in length. These fully martensitic specimens were then heated directly to the test temperature and immediately compressed.

4.2. Dilatation testing

The thermal volume change (dilatation) is the driving force for the thermal stresses induced in the material. It is the total effect of the thermal expansion and the volume change due to phase transformations. Because only martensitic transformation was required in the tests, a high cooling rate was necessary. To achieve this, a 10 mm in dia. rod with a length of 110 mm with a reduced middle section of 4 mm dia. $\times$ 5.5 mm in length was used in the dilatation testing. This specimen geometry enabled cooling rates of up to 280°C s^{-1} . However, for the dilatation tests the specimen was first fully austenitized at a temperature of 900°C for 5 min, followed by controlled cooling at a rate of 50°C s^{-1} . The dilatation was then obtained from the measured crosswise strain component.

5. Results and discussion

5.1. Flow stress curves of the austenite phase (900 – 650°C)

The strain hardening functions for the austenite phase at different temperatures with a constant strain rate of 0.1 s^{-1} , as seen in figure 2, describes the relative increase in yield stress due to true strain. These functions were obtained from the experimental longitudinal strain (corrected to account for Gleeble strain) and crosswise strain measured during the compression tests. These tests were performed to strains of 0.8–0.9 but due to significant bulging and non-uniform deformation of the test pieces, the data for strains beyond 0.5 should be considered unreliable. The curves show an initial work hardening, which becomes balanced by the dynamic recovery so that a steady state is eventually reached. For the 900°C curve it is even possible to detect the effects of dynamic recrystallization which is seen as a fall in the stress from a peak of around 150 MPa and peak strain of about 0.31. For comparison, the flow stress of a plain carbon steel (0.092%C–0.45%Mn) at $900^\circ\text{C}/0.1 \text{ s}^{-1}$ taken from a previous study [17], showed a peak stress

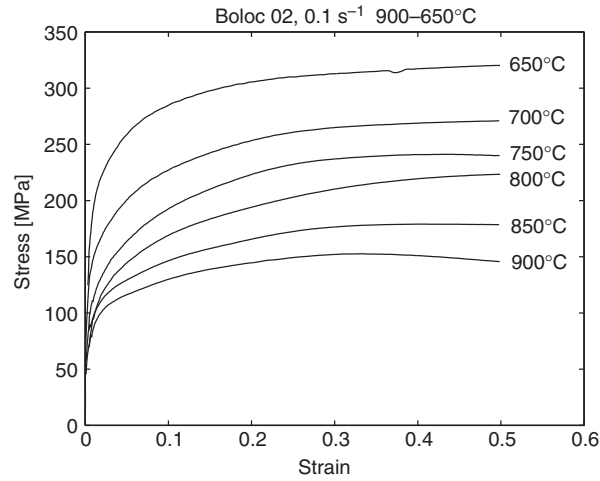


Figure 2. Stress–strain curves at temperatures of 900–650°C at a strain rate of 0.1 s⁻¹.

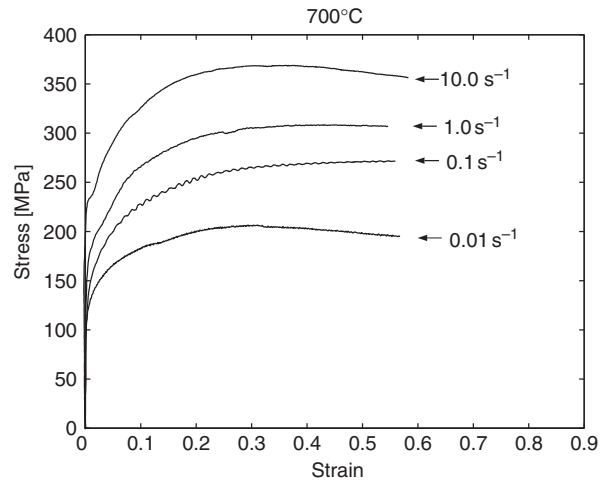


Figure 3. Influence of strain rate on flow curves at 700°C.

of 132 MPa. At a strain rate of 0.01 s⁻¹, the peak stresses are 110 and 106 MPa for the boron steel tested in the present paper and a plain carbon steel, respectively. This suggests that the effect of alloying elements (carbon, chromium and boron) on the flow stress level is relatively small (about 10%). It is commonly observed that the alloying has only a marginal influence on the flow stress level at high temperatures.

5.1.1. The influence of strain rate. The influence of strain rate in the austenitic phase was evaluated by high-temperature compression tests (900–650°C) at different strain rates. In figure 3 the flow stress curves at 700°C are shown for various strain rates. In general, it can be seen that the stress level increases with increasing strain rate. At the highest strain rates of

$5\text{--}10\text{ s}^{-1}$, adiabatic heating due to plastic deformation results in a decrease of the flow stress at high strains >0.4 (not dynamic recrystallization). One model (see [17]) describing the strain rate sensitivity of the austenite phase takes the relationship between the peak stress σ_p (MPa) (or a maximum stress at 0.4 strain in the instance of continuing work hardening) and the Zener–Hollomon parameter Z :

$$\sigma_p = a_0 \ln(Z) - a_1, \quad (5.1)$$

where the Zener–Hollomon parameter is given by

$$Z = \dot{\epsilon} \exp\left(\frac{Q_{\text{def}}}{RT}\right) \quad (5.2)$$

and $\dot{\epsilon}$ is strain rate, R the universal gas constant, T absolute temperature and Q_{def} is the activation energy for deformation. If a value 330 kJ mol^{-1} is taken for the activation energy, a value slightly higher than for C–Mn steels [17], the results for the austenite phase are $a_0 = 18.3$ and $a_1 = 424$, with the correlation factor $R^2 = 0.957$. This relationship was, however, not implemented in the presented material model and hence experimental strain rates comparable to the process being analysed were required.

5.2. Flow stress of phase mixtures (600–300°C)

The flow stress curves recorded in the intermediate temperature regime between 600 and 300°C, with a strain rate of 0.1, are shown in figure 4. These reveal a considerable increase in the flow stress level at $T < 425^\circ\text{C}$; apparently resulting from an increasing amount of the martensite phase being present. The shapes of the flow curves are also somewhat different in the range 425–500°C indicating an increasing work hardening rate at strains above 0.2. This is presumably the result of the formation of a harder phase in the deformed austenite. This might be deformation-induced martensite, formed above M_s or bainite formed due to the time spent at this temperature and accelerated transformation due to the concurrent deformation. From figure 1 it can be concluded that in this temperature regime (600–300°C) the cooling rate and

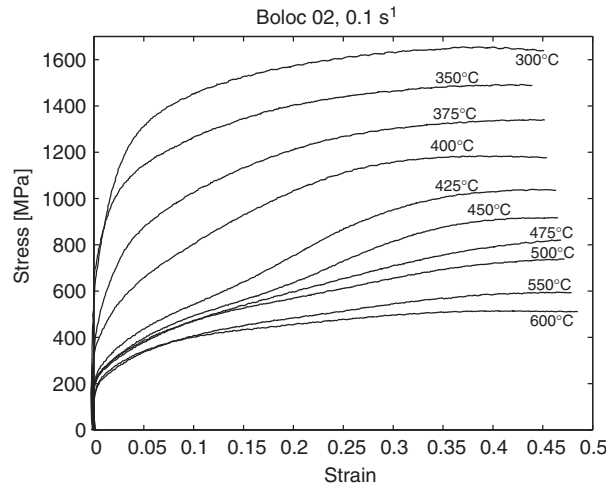


Figure 4. Stress–strain curves at temperatures 600–300°C and a strain rate of 0.1 s^{-1} .

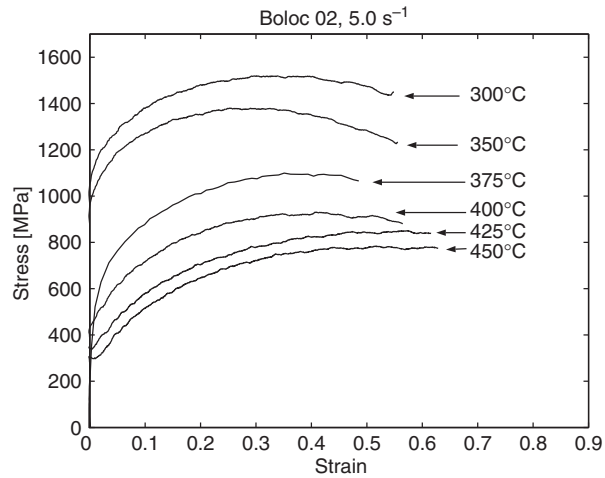


Figure 5. Curves for stress versus strain at temperatures 450–300°C at a strain rate of 5.0 s^{-1} .

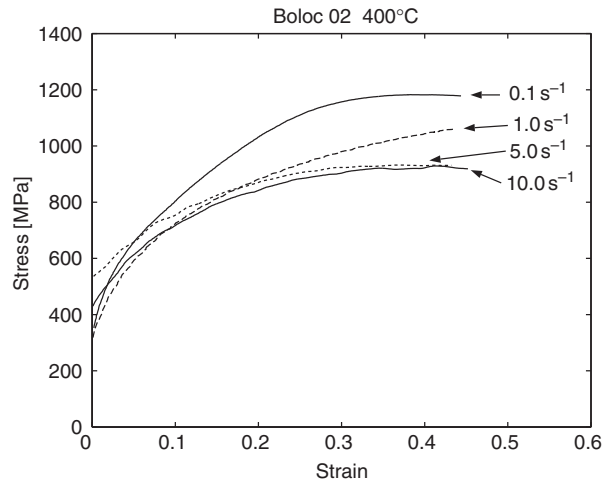


Figure 6. Influence of strain rate on flow stress at 400°C.

holding time before testing can have a significant influence on the microstructure of the steel and consequently on its stress–strain behaviour. This is due to transformation of austenite into bainite, which starts quite quickly, within 20 s at 450°C; some dilatation tests indicated an even shorter time. To minimize bainite transformation in the initial portions of the stress–strain curves, some compression tests with a strain rate of 5.0 s^{-1} were performed, the results being shown in figure 5.

5.2.1. The influence of strain rate. The influence of strain rate was evaluated at temperatures of 400 and 500°C; the results being shown in figures 6 and 7. The applied strain rates were 0.1, 1.0 and 5.0 or 10.0. In this temperature range, bainite formation occurs very rapidly.

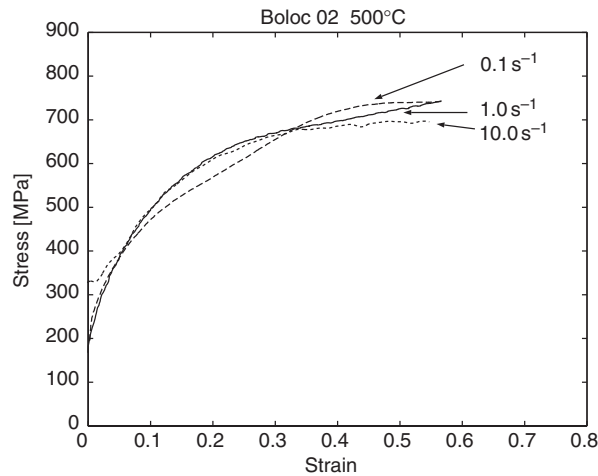


Figure 7. Influence of strain rate on flow stress at 500°C.

The greatest cooling rate achievable was approximately $50^{\circ}\text{C s}^{-1}$ from 900°C , and decreases gradually at lower temperatures. From this it is clear that bainite formation is inevitable at these temperatures. It is also known that deformation may accelerate austenite decomposition. At higher strain rates, bainite transformation may be largely avoided in the initial portions of the stress–strain curves due to less time spent in that temperature regime. Therefore, the influence of strain rate may become even opposite to that at high temperatures, while more bainite will be formed at a lower strain rate than at a higher one. For instance, at 400°C , if strain rates less than 1.0 are applied, as seen in figure 6, there will be a significant increase in flow stress, presumably due to a transformed bainite phase. This is less apparent at 500°C . An additional factor, which cannot be ignored, is the effect of adiabatic heating due to plastic dissipation whilst martensitic transformation induced by plastic deformation may also affect the behaviour of the material during processing.

5.2.2. Evaluated flow stress curves of the intermediate phase ($600\text{--}300^{\circ}\text{C}$). The hardening law is defined by tabulated values of yield stress versus effective plastic strain. The flow stress curves of the intermediate phase presented in figure 8 thus describes the relative increase in yield stress due to true plastic strain. During compression at intermediate temperatures ($600\text{--}300^{\circ}\text{C}$), it was found that the time available for the material samples to cool from austenitization temperature to the actual testing temperature and to perform the test was insufficient to avoid bainite formation. As the influence of strain rate is less apparent at the intermediate temperatures ($600\text{--}300^{\circ}\text{C}$) than in the austenitic range ($900\text{--}650^{\circ}\text{C}$), it is possible to obtain stress–strain curves less affected by bainite phase transformations by performing the compression tests at elevated strain rates when the temperature is close to M_s . In figure 8, the flow stress curves at strain rates of 0.1 and 5.0 s^{-1} (for the tests at 400 , 425 and 450°C) are shown. The equivalent curves for these lower temperatures at a strain rate of 0.1 s^{-1} are shown in figure 4 and reveal a markedly higher stress level at this lower strain rate. It is important to remember that the specimen temperature tends to increase due to adiabatic heating at higher strain rates, especially as the hard martensite phase is present, making proper testing difficult.

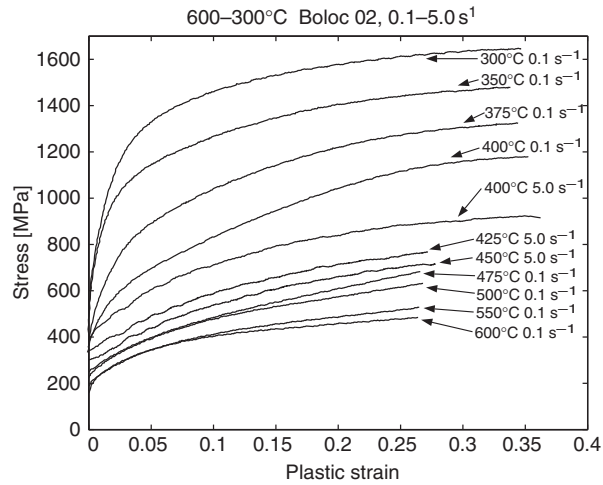


Figure 8. Evaluated curves of stress versus plastic strain at temperatures 600–300°C.

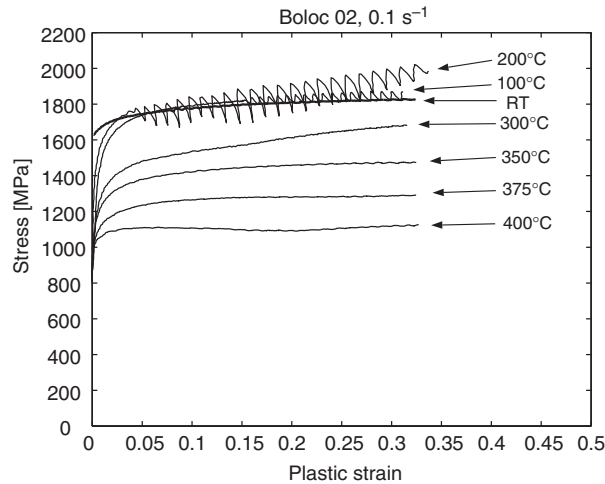


Figure 9. Stress–plastic strain curves at temperatures 400–20°C.

5.3. Flow stress of the martensite phase (400–20°C)

In the full martensite phase, 400°C to RT (20°C), the flow stress curves are relatively flat, i.e. work hardening in martensite is weak, but the stress level is very high; above 1600 MPa at 300°C, see figure 9. Some specimen showed non-uniform deformation and buckling when the plastic strain exceeded 0.3 and only the data for strains below this level should be considered reliable. A pronounced ‘jerky’ flow can be seen at 100–200°C, presumably caused by dynamic interaction between carbon atoms and moving dislocations and the flow stress increases above that at RT. The curves shown in figure 9 for the completely martensitic structure (300–400°C) were not used in the FE modelling. The curves presented in figure 8 are assumed to be more representative of the material flow stress behaviour during quenching.

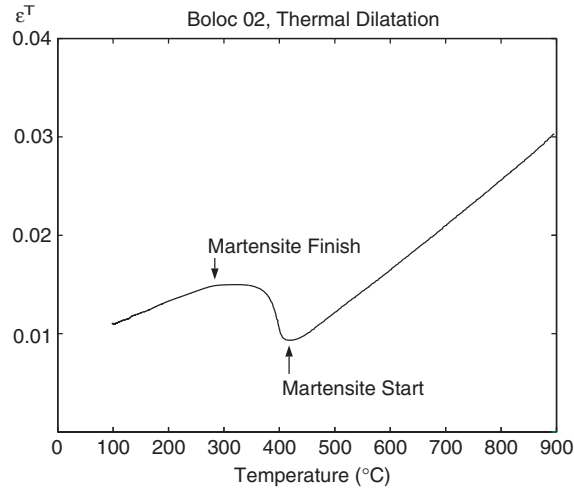


Figure 10. Thermal dilatation as a function of temperature.

5.4. Dilatometric test results

The thermal volume change (dilatation) was obtained from the measured crosswise strain component and is shown in figure 10. This shows that martensitic transformation starts at approximately 420°C and finishes by approximately 280°C.

6. Modelling examples

6.1. Tubular beam subjected to high-temperature forming

The results from one example of a forming simulation, using the experimentally obtained material data, are presented below. This example assumes a constant forming temperature of 800°C and hence an isotropic elastic–plastic material model could be used (see also [18]). The initial yield stress was 59 MPa and the elastic material parameters and friction coefficients (see table 2) used were obtained from the experiments and literature [11, 16]. The values of effective stress ($\bar{\sigma}$) versus plastic strain (ϵ^p), used to describe the piecewise linear hardening function, are given in table 3. These values, corrected for the elastic portion, were obtained from the flow stress curve at 800°C with a strain rate of 0.1 s^{-1} , (see figure 2).

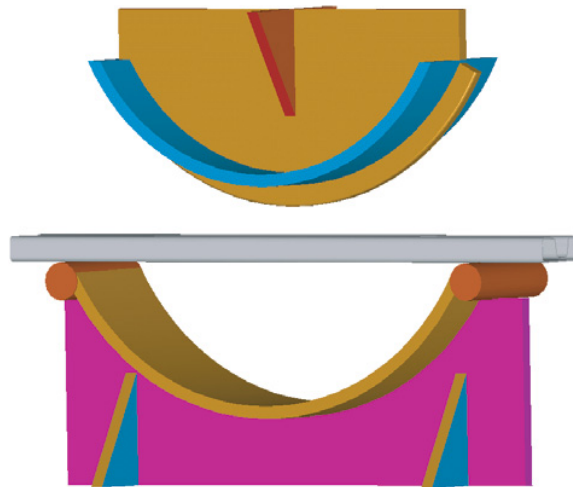
The forming process modelled and shown in figure 11 is the high-temperature bending of a thin-walled closed section steel profile. The forming tool consists of two parts where the upper part has an outer radius of 0.18 m and the lower part an inner radius of 0.2 m, as described in [18]. The geometry model was created using the IDEAS CAD system, and the explicit finite element code DYNA3D was used for the forming analyses (see also [19]). The DYNA3D input files (FE models) were generated using the integrated system described in [20] and displayed as one half of the actual geometry (see figure 12). The experimental press force was measured using a force transducer attached to the upper forming tool. This force is compared to the calculated pressing force from analysis (curve with asterisk) and shows good agreement in the early stages of the forming process, as displayed in figure 12. In the later stages of the process there is a deviation due to cooling and local hardening in the processed beam. This is, however, not accounted for in this isothermal model.

Table 2. Material parameters.

Parameter	Value
Young's modulus	76 GPa
Poisson's ratio	0.31
Initial yield stress	59 MPa
Static friction coefficient	0.2
Dynamic friction coefficient	0.2

Table 3. Effective von Mises stress ($\bar{\sigma}$) versus plastic strain (ϵ^P).

(MPa)	Value
58.61	0.000
91.25	0.004 37 3
115.0	0.014 23
138.3	0.035 32
163.9	0.081 64
187.6	0.1623
209.7	0.2905
332.3	1.000

**Figure 11.** Forming tool and the boron steel profile.

6.2. Compression of a cylindrical specimen during continuous cooling

A second example comparing experimental data with the results from the numerical analysis concerns the compression of a cylindrical test specimen during continuous cooling. Here the results of the numerical analysis are compared with actual resulting force and radial displacement measured in a Gleeble experiment. In the experiment, compression started at 800°C with a strain rate of 0.064 s⁻¹ and ended after 8.5 s with a final temperature of approximately 400°C. To obtain the required cooling rate, over 40°C s⁻¹, a specimen with a diameter of 5.0 mm and a height of 7 mm was used. Specimen symmetry allowed a finite element model of one fourth of the sample geometry to be used. The specimen and the anvil,

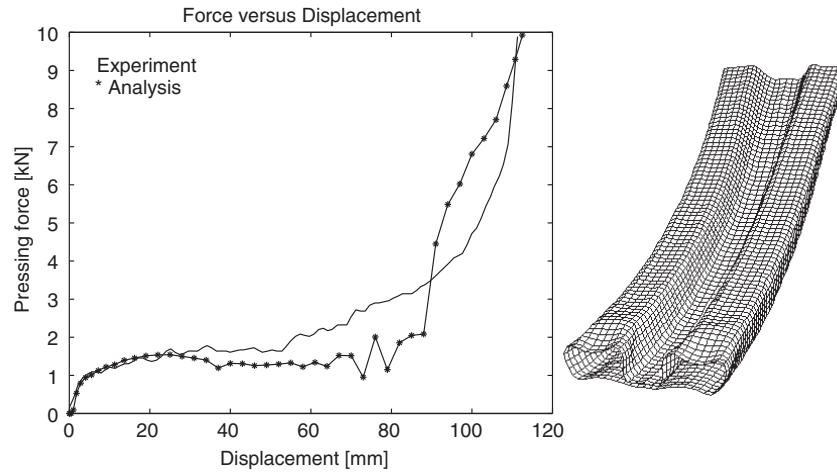


Figure 12. Plots of measured and calculated (curve with ‘*’) forces during forming at 800°C.

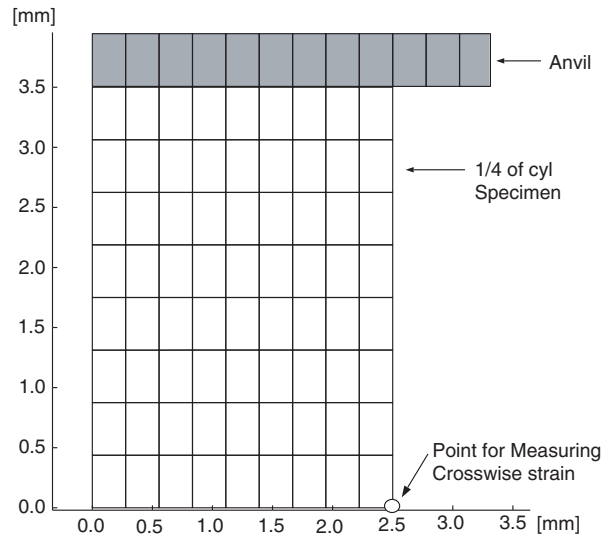


Figure 13. Finite element model of the Gleeble experiment.

shown in figures 13 and 17, consist of 84 four-node axisymmetric quadrilateral elements. The motion of the anvil is controlled by defined displacement boundary conditions. To enable comparison with experimental results the actual experimental anvil velocity profile was used. In this analysis, an elastic–plastic material model with a tabulated hardening function was used. The temperature-dependent mechanical properties were defined for specific temperatures with intermediate values obtained by interpolation. The temperature field was assumed to be uniform with the temperature history taken from experimental data. The tabulated values of yield stress versus true plastic strain were taken from actual measurements and are shown in figure 14 as curves. The elastic behaviour of the material is described by Young’s modulus and Poisson’s ratio. Poisson’s ratio was assumed to be constant, 0.3, whilst Young’s modulus was

assumed to vary according to table 4. The analysis also took into account the thermal expansion function, which was extracted from the dilatation curve presented in figure 10, which included both thermal expansion and volume change due to phase transformation.

The results, presented in figures 15 and 16, show the measured and calculated axial forces and radial displacement. The radial displacement was measured at the mid-section of the

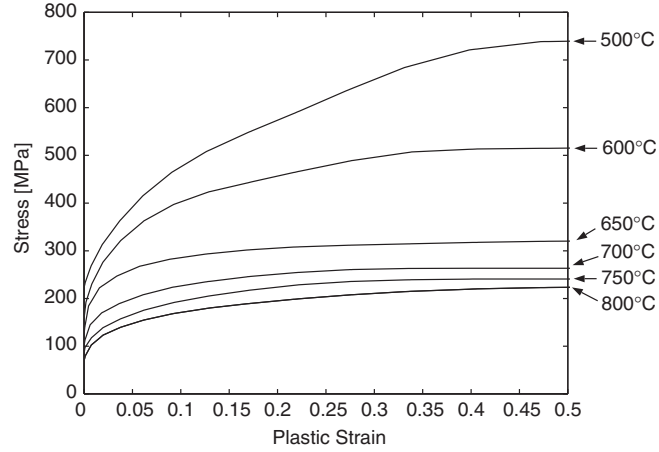


Figure 14. Stress versus strain at different temperatures derived from experiments.

Table 4. Young's modulus as a function of temperature.

Young's modulus (Pa)	Temperature (°C)
2.000E+11	20
1.400E+11	500
5.500E+10	900

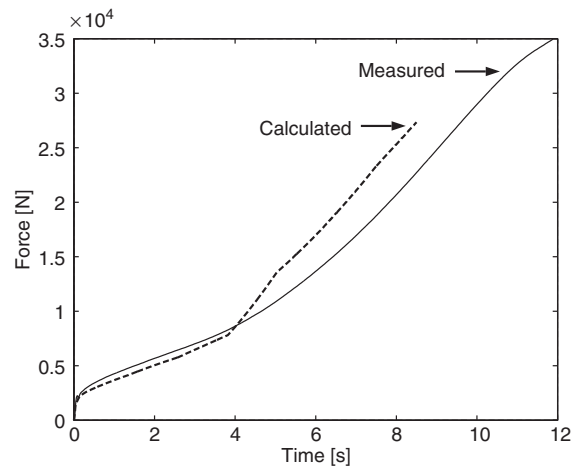


Figure 15. Plots of measured and calculated axial forces during compression.

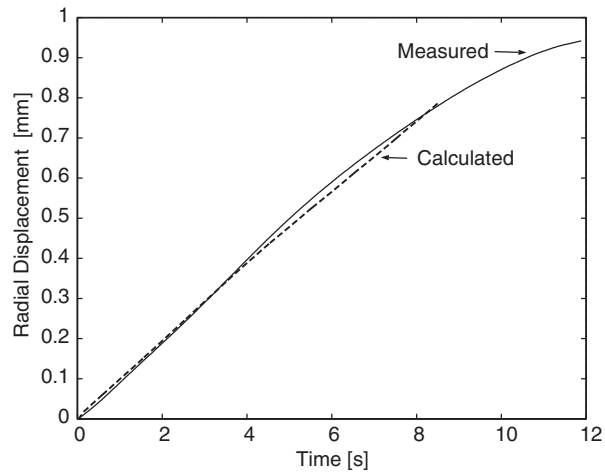


Figure 16. Plots of measured and calculated radial displacement during compression.

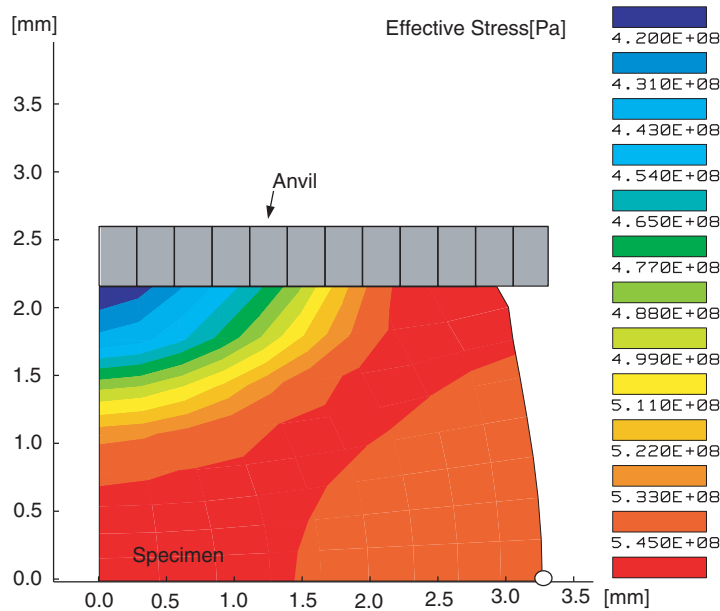


Figure 17. Colour contours of effective von Mises stress at 7.48 s into the compression analysis.

specimen (see figures 13 and 17). The predicted force shows acceptable agreement with the measured force for the first 4 s when temperatures were above 600°C. At temperatures below 600°C, the FE model seems to overestimate the yield stresses and resulting press force. This may be due to the material model not including dynamic recovery and the high plastic straining, occurring in the austenite phase, may hence result in overestimated yield stresses in the intermediate temperature regime (see figures 15 and 17).

7. Discussion and conclusions

The aim of this paper was to determine the mechanical material data necessary to allow numerical modelling of a combined forming and hardening process for thin-walled steel profiles involving rapid cooling and a martensitic phase transformation. In order to simulate this manufacturing process, there is a need for a thermo-elasto-plastic material model with a temperature-dependent non-linear hardening function. The uni-axial temperature-dependent stress-strain relationships and thermal dilatation were established by testing using a Gleeble 1500 thermo-mechanical simulator. A testing procedure was developed to allow isothermal compression tests to be carried out in a manner that circumvents the problems due to the short time available for the present material to cool from the high austenite temperature to the required testing temperature and to perform the tests. However, to obtain stress-strain curves unaffected by bainitic transformation, the compression tests had to be performed at elevated strain rates where the temperature was close to M_s . The use of still higher strain rates was found to result in problems with temperature control during the test. This may be due to energy dissipation during plastic deformation or to latent heat during martensitic transformation. At temperatures above 600°C, the strain rate sensitivity of the flow stress was found to be considerable due to the influence of dynamic recovery, but at lower temperatures the direct influence of the deformation rate was small and the flow stress appeared to be related to the time-dependent decomposition of austenite. The stress-strain curves in this interval may thus be taken from tests carried out at higher strain rates, which appear to be less affected by bainitic transformations. A complete set of stress-strain curves in axisymmetric compression in the temperature range from 900°C down to RT was obtained. This material data was used in two finite element simulations; a high-temperature forming assuming isothermal conditions and a compression experiment during continuous cooling. For these simulations, the material models used, which were based on experimental data, were found to give acceptable agreement with the experimental results.

In future work this mechanical model is intended to be used together with a non-linear thermal model (see [12]), which requires the following thermal material parameters to be determined.

- (a) Transformation plasticity coefficient $(\Delta V/V)(1/\sigma_y)$. The transformation plasticity (see [12]) is an irreversible deformation that occurs when a material undergoes phase transformation under applied stresses.
- (b) Thermal conductivity as a function of temperature. The thermal conductivity function is needed for a complete thermo-mechanical analysis.
- (c) Heat capacity as a function of temperature. The heat capacity function includes the effects of the latent heat released during the transformation from austenite to martensite.

Future work will also focus on improving the material model by implementing dynamic recovery, transformation plasticity, etc. The estimation of material parameters at elevated temperatures may also be improved by using inverse modelling techniques allowing non-uniform strain, stress and temperature fields to be accounted for.

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