Modeling the Effect of Solute Drag on Recovery and Recrystallization during Hot Deformation of Nb Microalloyed Steels

Linda BÄCKE

Dalarna University, Borlänge, Sweden/SSAB Strip Products, Borlänge, Sweden. E-mail: linda.backe@ssab.com

(Received on March 11, 2009; accepted on November 22, 2009)

The effect of solute drag on recovery and recrystallization during hot deformation of Nb microalloyed steels has been modeled using a newly developed microstructure model. The model is based on dislocation theory and the calculated dislocation density determines the driving force for recrystallization. Subgrains act as nuclei for recrystallization and have to reach a critical size and configuration in order for recrystallization to start. In the model, the solute drag effect of Nb in solution is described. Nb retards both dislocation and grain boundary movement giving retardation in both recovery and recrystallization. Calculations were compared to experimental results from axisymmetric compression tests combined with stress relaxation. In order to model the effect of solute drag, the experiments were carried out at temperatures where precipitation of Nb(C, N) should not occur. The calculated flow stresses for the compression tests show good fit with experimental data. Also, the calculated results of the relaxation tests show good agreement with experimental data.

KEY WORDS: modeling; flow stress; recrystallization; recovery; solute drag; Nb microalloyed steel.

1. Introduction

Nb is known to give the largest retardation of recrystallization.1) The retarding effect depends both on Nb in solid solution and on precipitated niobium carbo-nitrides.2,3) It is the impeding effect of Nb on the movement of austenite grain boundaries and subgrain boundaries that causes the retardation of recrystallization.

The effect of solute drag on recovery and recrystallization has been investigated for years.2–10) According to Yamamoto,3) a plausible explanation to the extent of the retarding effect of solutes on recovery and recrystallization is the solutes lattice distortion. He found that Nb retards recrystallization and Mn only gives a small retardation. In this investigation, only Nb is considered and the retarding effect of other elements, such as Mn, Si and Al is assumed negligible.

This investigation is a continuation of the work by Engberg and Lissel11) where a physically-based model for describing the microstructural evolution of austenite during hot deformation was validated for a plain C–Mn steel. In the present paper, the solute drag effect of Nb on recovery and recrystallization is investigated using the microstructure model.

The solute drag effect on recrystallization has been modeled with several approaches during the past decades.4,5,8–10) Medina and Mancilla8) found that for Nb, Ti, Mn and Si, the apparent activation energy used in empirical models describing recrystallization kinetics increased. Suehiro and Senuma9) used a model first developed for phase transformation, to investigate the effect of solute drag on recrystallization. Zurob et al.10) used a simplified model by Cahn11) to describe the solute drag effect. In a previous investigation by Lissel et al.,12) the model by Cahn was used to describe the solute drag effect during hot rolling of Nb steel. This gave satisfying results compared to measured data. The simplified model is only valid for low grain boundary velocities, i.e. not the initial stage of recrystallization when the driving force is large. Hillert and Sundman5) developed a model for solute drag on moving grain boundaries and phase interfaces and provided an analytical solution that gives a description for the retarding force at all velocities. Their analytical solution is easy to combine with the microstructure model. Hence, in this investigation, this model was chosen for its more adequate description of the solute drag effect.

2. Experimental Procedure

Two commercial HSLA steels with chemical compositions presented in Table 1 were used in this investigation. The steels were collected from the hot strip mill at SSAB Strip Products Division. Cylindrical specimens with height of 12 mm and diameter of 10 mm were machined from the head end of transfer bars after rolling in the roughing mill. In the reheating furnaces at the hot strip mill the slabs are heated to about 1250°C and in order to simulate the conditions for the hot strip mill, the samples were pre-heated at
Table 1. Chemical composition of the steels (mass%).

| Steel | C   | Si  | Mn  | P   | S   | N   | Al  | Nb | V  | Ti |
|-------|-----|-----|-----|-----|-----|-----|-----|-----|----|----|
| Nb1   | 0.065 | 0.007 | 1.085 | 0.014 | 0.003 | 0.004 | 0.055 | 0.034 | 0.006 | 0.001 |
| Nb2   | 0.066 | 0.036 | 1.323 | 0.011 | 0.003 | 0.005 | 0.040 | 0.055 | 0.005 | 0.001 |

1 250°C and soaked for 40 min. Immediately after the solution treatment, the samples were water quenched. The solution treatment of the samples had two purposes, eliminating the rolling texture and dissolving Nb(C, N) particles.

The temperature selected is well above the equilibrium solution temperature calculated with Thermo-Calc and the database TCFE5, see Fig. 1. For the Nb1 steel, the solution temperature for Nb(C, N) was about 1050°C and it was about 1100°C for the Nb2 steel. Also AlN is taken into consideration in the Thermo-Calc calculations, which is not shown in the figure although it influences the amount of Nb(C, N).

Compression tests and stress relaxation tests were carried out in order to determine the effect of Nb on strengthening and on recrystallization kinetics. Stress relaxation tests enable to obtain a complete recrystallization curve from only one test and the technique is proven to be a reliable method for characterizing both static and metadynamic recrystallization processes. The tests were performed on a Gleeble 1500 thermomechanical simulator at Oulu University, Finland. The details regarding the experimental procedure is described elsewhere.

The compression tests were performed up to a true strain of 0.7 at temperatures 1000, 1100 and 1200°C at a strain rate of 10 s⁻¹. The stress relaxation tests were performed at temperatures between 1000°C and 1200°C, at pre-strains of 0.2, 0.4 and 0.6 and strain rates of 0.1 and 1.0 s⁻¹. Before the deformations, all samples were first heated to 1200°C with a heating rate of 4°C/s, soaked for 3 min, and then cooled to the deformation temperature with a cooling rate of 10°C/s. To remove the temperature gradients in the samples, they were soaked at the desired deformation temperature for 15 s prior to the deformation.

The size of the austenite grains prior to deformation was measured on a sample heated to 1200°C, soaked for 3 min and then water quenched. The measurements were made at Oulu University.

3. The Model

A model used for calculating the microstructural evolution has been developed by Engberg and is available in a toolbox for process simulation programmed in MATLAB. It is based on the work by Siwecki and Engberg and Wang et al. and is built up by several sub-models describing dislocation generation and recovery, creation and diffusion of vacancies, formation and growth of subgrains, recrystallization and grain growth.

The model is described in detail and validated in a work by Engberg and Lissel. It was shown that the model can predict the flow stress behavior of a plain C–Mn steel during hot compression tests combined with relaxation. This model can describe the microstructural evolution during hot rolling of microalloyed steels. Some minor changes have been done in the model and also models for the solute drag effect on grain growth and recovery are added. Below follows a short description of the model, where the effect of solute drag is presented.

3.1. Work Hardening and Recovery

The flow stress in the model is described by the established equation

\[ \sigma_i = \sigma_0 + m \cdot \alpha \cdot G \cdot b \cdot \sqrt{\rho} \]

which describes the flow stress well, see for example the work by Bergström. The first term on the right side, \( \sigma_0 \), consists mainly of the contribution of precipitates. In the second term on the right side, \( m \) is the strain dependent Taylor factor (~3.1), \( \alpha \) is a constant (~0.15), \( b \) is the Burgers vector (~2.58 Å) and \( G \) is the temperature dependent shear modulus. The index, \( i \), indicates if the flow stress and dislocation density, \( \rho \), is calculated for deformed or recrystallized material.

During hot deformation there is a substantial amount of recovery simultaneous with work hardening. In the model, the recovery is described by both glide and climb. The climb process is dependent on the diffusion of vacancies and is given by

\[ \frac{dP_{\text{climb}}}{dt} = -M_m \cdot C_v \cdot \rho^2 \cdot \left( 1 - \frac{3}{4} \frac{f_r}{r} \right) \]

where \( M_m \) is the rate parameter of dislocation generation and \( C_v \) is the vacancy concentration. The second term in the parentheses describes the pinning effect on the recovery due to particles where \( r \) is the mean radius of second-phase particles and \( f_r \) is their volume fraction. Equation (2) is developed to give a good description of the recovery in the initial stage subsequent to deformation. Hence, the recovery is well described in processes with rather short holding.
times, as in hot strip rolling.

Apart from the retardation of recovery due to precipitates, the recovery is also retarded by Nb in solution\(^{1,6,7}\) and therefore, the rate parameter of dislocation recovery is reduced. This is described by\(^{6}\)

\[
M_m = \frac{M_{m0}}{1 + \left( \frac{k_mD_x^{x_m}}{D_mT^2} \right) s} \tag{3}
\]

where \(M_{m0}\) is the rate parameter without solute drag, \(k_m\) is a constant (~\(1 \times 10^{10}\)), \(D\) is the diffusion coefficient for the element, \(x_m\) is the solute concentration, \(T\) is the temperature, and \(D_m\) is the self diffusion coefficient of the matrix. The diffusion coefficients used in the calculations are taken from Fridberg et al.\(^{20}\) and Oikawa\(^{21}\) and are given in Table 2.

### Table 2. Diffusion coefficients used in the calculations.

|     | Fe | Nb |
|-----|----|----|
| \(D_s\) [m/s] | \(7 \times 10^{-7}\) | \(5.6 \times 10^{-4}\) |
| \(Q\) [J/mol] | 286000 | 286000 |
| Reference | 20\(^{9}\) | 21\(^{10}\) |

#### 3.2. Subgrain Formation

Due to recovery, subgrains are formed during deformation. When more dislocations are generated, the subgrain size, \(R_{sub}\), is reduced. In the model it is assumed that the subgrain size is proportional to the mean free distance of dislocation slip and the subgrain development is described by\(^{11}\)

\[
dR_{sub} \frac{dt}{\rho} = -\frac{k_i}{2\sqrt{\rho}} \left( \frac{d\rho}{dt} \right) \frac{R_{sub}^2}{\rho} \tag{4}
\]

where the parameter \(k_i\) is a constant.

The boundary energy of subgrains, \(\gamma_{sub}\), is assumed to be dependent on the boundary energy of a high angle grain boundary, \(\gamma_{gr}\) (~0.8 J/m\(^2\))

\[
\gamma_{sub} = \gamma_{gr} \frac{\theta_{sub}}{\theta_{gr}} \left( 1 - \ln \frac{\theta_{sub}}{\theta_{gr}} \right) \tag{5}
\]

where \(\theta_{sub}\) and \(\theta_{gr}\) are the misorientation of low and high angle grain boundaries, respectively. The size of the subgrains is dependent on the dislocation density and it is also assumed that the misorientation of subgrains is dependent on the dislocation density. The ratio in Eq. (5) between the misorientation of subgrains and that of high angle boundaries is therefore assumed to be dependent on the recovery of dislocations.\(^{11}\) It is assumed that the misorientation of the subgrains increases as the deviation between dislocation density, with and without recovery, increases. This is described by the ratio between the actual dislocation density, \(\rho\), with recovery and the dislocation density generated without recovery, \(\rho_{gen}\), and is described by

\[
\frac{\theta_{sub}}{\theta_{gr}} = 1 - \frac{\rho}{\rho_{gen}} \tag{6}
\]

A critical subgrain size and configuration determine the onset of recrystallization.

#### 3.3. Recrystallization

When recrystallization has been triggered, the recrystallizing grains grow into the deformed grains, replacing the deformed material. Recrystallization is driven by the stored energy of deformation, i.e. it depends on the change in dislocation density as the recrystallization front moves. The driving force is reduced by the retarding forces due to Zener pinning, \(P_Z\), and by solute drag, \(P_S\), and is described by

\[
F_{rec} = -\frac{\gamma_{gr} c_d G h^2}{R_{rec} \rho_{def} - \rho_{rec}} - P_Z - P_S
\]

\tag{7}

where \(R_{rec}\) is the mean size of recrystallized grains, \(c_d\) is a constant (0.5) and \(\rho_{def}\) and \(\rho_{rec}\) is the dislocation density in the deformed and recrystallized material, respectively. The retarding force due to Zener pinning is for randomly distributed particles described by\(^{23}\)

\[
P_Z = k_m \frac{f_v}{r} \tag{8}
\]

where \(k_m\) is a constant (~\(3\gamma_{gr}/2\)). The retarding force due to solute drag is here considered for Nb. This force is described by the model by Hillert and Sundman\(^{5}\)

\[
P_S = \frac{RTv}{V_m} \int_y \frac{\gamma_{sl} x_{sl}^m}{D(1-x_{sl}^m)x_{sl}^g} \tag{9}
\]

where \(R\) is the gas constant (8.314 J/mol K), \(v\) is the velocity of the grain boundary, \(V_m\) is the temperature dependent molar volume, with values taken from Thermo-Calc,\(^{13}\) \(\gamma_{sl}\) is the mole fraction of the solute in the grain boundary, \(x_{sl}^m\) is the solute concentration and \(v\) is the distance from the boundary. The concentration profile of the solute across the grain boundary is dependent on the interaction energy, \(\Delta G\), between the solute and the grain boundary. The variation of the interaction energy is described by a truncated wedge-shaped function, see Fig. 2. This gives a constant value of \(d\Delta G/\text{dy}\) with opposite sign at the two sides of the boundary and zero in the middle of the boundary. The diffusion coef-

![Fig. 2. Truncated wedge-shaped function of the interaction energy \(\Delta G\) across the boundary.](image-url)
sufficient is assumed to be constant. The concentration profile over the boundary is then described by

\[
\frac{dx_{ag}}{dy} = -\frac{1}{RT} \frac{d\Delta G}{dy} \left( (1 - x_{ag}) \right) - \frac{V}{D} (x_{ag} - x_{ag}^\infty)
\]

for which the analytical solution provided by Hillert and Sundman has been used in the model.

For recrystallization to start, the ratio between the misorientation of subgrains and grains, Eq. (6), has to be close to one (0.84 was found to be a good value) and the subgrain size has to be larger than a critical value. When recrystallization starts, the size of recrystallized grains is equal to the size of the subgrains and the critical subgrain size is calculated from Eq. (7) with the driving force of recrystallization equal to zero. Hence, the following conditions has to be fulfilled

- \( \theta_{sub} \theta_{gr} \geq 0.84 \)
- \( R_{sub} \geq \frac{\gamma_{gr} - \gamma_{sub}}{c_d G \gamma h^2 (\rho_{def} - \rho_{def})} \cdot F_{rec} \)

The size of recrystallized grains, \( R_{rec} \), in Eq. (7) is calculated from both the size of the nuclei of recrystallization and the size of the growing recrystallized grains, \( R_{recg} \).

The growth rate of recrystallizing grains are dependent on the mean value of the vacancy concentration on both sides of the moving boundary and is described by

\[
\frac{dR_{rec}}{dt} = k_r \cdot M_g \cdot 0.5 \cdot (c_{v,def} + c_{v,rec}) \cdot F_{rec} \]

where \( k_r \) (\( \sim 1 \)) is a constant, \( M_g \) is the grain boundary mobility and \( c_{v,def} \) and \( c_{v,rec} \) are the vacancy concentration in the deformed and recrystallized material, respectively.

The fraction recrystallized, \( X_{rec} \), is calculated from the size of recrystallized grains. It is assumed that all grains have the same volume. The fraction recrystallized is the number of recrystallized grains times the volume of a recrystallized grain.

The rate of recrystallization is hence described by

\[
\frac{dX_{rec}}{dt} = c_f \left( \frac{dN_{rec}}{dt} \cdot R_{rec}^3 + N_{rec} \cdot 3 \cdot R_{rec}^2 \cdot \frac{dR_{rec}}{dt} \right)
\]

where \( c_f \) is a constant (\( \approx 4\pi/3 \) for spherical grains) and \( N_{rec} \) is the nucleation sites.\(^{11)}\)

Once the recrystallizing grains impinge and the grains grow on the expense of each other, the growth is described by normal grain growth

\[
\frac{dR_{grow}}{dt} = k_g \cdot M_g \cdot c_{v,rec} \cdot F_{grow} \]

where \( k_g \) is a constant (\( \sim 1 \)) and \( F_{grow} \) is the driving force, described by

\[
F_{grow} = \frac{\gamma_{gr}}{R_{rec}} - P_z - P_S \]

4. Results and Discussion

The two steels were given the same dissolution and heat treatment before deformation. This resulted in different grain sizes, the Nb1 steel obtained a grain size of 260 \( \mu m \) and the Nb2 steel a grain size of 160 \( \mu m \) after a 3 min soak at 1200°C.

The volume fraction of possible second-phase particles which can retard grain growth is shown in Fig. 1. At 1200°C, no Nb(C, N) should be present and the only particles exhibiting Zener pinning on the grain growth is MnS. According to the equilibrium calculations using Thermo-Calc, the volume fraction of MnS at 1200°C was 8.7 \( \times 10^{-5} \) and 9.0 \( \times 10^{-5} \) in the Nb1 and Nb2 steel, respectively. According to Hayashida et al.,\(^{24)}\) the mean size of the MnS after vacuum melting and casting is around 0.05 \( \mu m \) for a steel with about the same amount of sulfur as the steels in the present investigation. In steels with higher content of sulfur (0.015 mass%), coarser particles with size 0.2 \( \mu m \) where found. Avdusinovic and Gigovic\(^{25)}\) found that the mean size was about 0.8 \( \mu m \) at 1200°C for a steel with 0.015 mass% sulfur. In the calculations, the mean size of MnS was set to 0.1 \( \mu m \) for the two Nb steels.

For the Nb steels, the selected solution temperature is well above the equilibrium solution temperature of Nb(C, N) and it is not likely that any Nb(C, N) particles remained undissolved after this heat treatment. Compared to the measured grain size for a C–Mn steel,\(^{11)}\) with more volume fraction of MnS, heated at the same temperature, the grain size of the Nb1 steel is smaller, see Fig. 3. Hence, the difference in grain size depends on a solute drag effect due to Nb in solution. The grain size in the Nb2 steel is smaller than in the Nb1 steel and with no Nb(C, N) particles left undissolved. This should be an effect of the higher Nb content in solution. The effect of solute drag is described by Eq. (9) and the adjustment parameters are the grain boundary width and the interaction energy. With a grain boundary width of 0.5 nm the interaction energy that provided the best fit between the calculated grain sizes and the measured ones was \(-13\ 500\) J/mol.

As seen in Fig. 3, the difference in measured grain size between the two Nb steels can not be exactly described with the solute drag model. The calculated grain size for the Nb2 steel is somewhat higher than the measured value. Increasing the value of the interaction energy would give a better description of the measured grain sizes but it would also give a larger retardation of recrystallization at 1200°C.
which is not the case as will be evident in the results below. However, since the calculated values are in reasonable agreement with measured values the model is thought to describe the grain growth adequately. In the calculations below, the measured grain size is used as input data to the model.

No precipitation of Nb(C, N) is calculated in this investigation, only the effect of Nb in solution is accounted for. PTT diagrams produced by Weiss and Jonas\(^2\) indicates that strain induced precipitation of Nb(C, N) in a 0.05%C–0.42%Mn–0.004%N–0.035%Nb steel at 1000°C starts after more than 100 s. Weiss and Jonas\(^2\) also studied dynamic precipitation, \emph{i.e.} precipitation during deformation. For dynamic precipitation, which is the case for relaxation, it appears that precipitation occurs somewhat earlier. This means that precipitation could occur in the latter part of the 1000 s long relaxation test at 1000°C. For the Nb2 steel there could be some precipitation during the early stages of the relaxation test at 1000°C.

In Fig. 4 to Fig. 6 flow stress curves from compression tests of the two Nb steels and a C–Mn steel\(^1\) are shown.

When comparing the results, the effect of Nb in solid solution on the hardening is evident. Recovery is reduced when more alloying content is added giving a higher flow stress. It can be seen that at low strain rates there is only a slight difference in the flow curves between the two Nb steels. However, for the Nb2 steel the flow curves do seem to have a trend of exceeding those of the Nb1 steel at larger strains for all strain rates. In addition, with increasing alloying content, the peak strain is shifted to higher values.

Akben \textit{et al.}\(^7\) proposed a way to study the recrystallization by constructing dynamic recrystallization-time–temperature (RTT) curves. The recrystallization start time at each temperature is taken as the peak strain through the strain rate. RTT curves were here constructed from the flow stress curves at strain rate 0.1 s\(^{-1}\) and temperature 1000–1200°C where a peak strain could be determined, see Fig. 7. Also included in Fig. 7 are the recrystallization times for the C–Mn steel. It can be seen that recrystallization is much faster in the plain C–Mn steel compared to the microalloyed steels. Though the difference in the time to peak stress is rather small between the microalloyed steels, there is a tendency for longer times for the Nb2 steel compared to the Nb1 steel. Even at 1200°C there is a solute drag effect on recrystallization, as it was for normal grain growth. The difference between the two Nb steels increases at lower temperature.

An example of the result from the stress relaxation is shown in Fig. 8. This figure shows that during relaxation there are significant fluctuations of the flow stress level. This is due to the hydraulic controller of the machine and the abrupt drop in stress after about 0.2 s is explained by Karjalainen \textit{et al.}\(^2\) as a machine characteristic. To smoothen the irregularities in the tests, in order to get good input data to the model, mean values of the strain, strain...
rate and temperature were taken over about 35 point intervals. The figure also reveals that there is somewhat slower recrystallization kinetics with higher alloying content.

4.1. Modeling Recrystallization during and after Deformation

In a previous investigation,11) the agreement between calculated and experimental flow stress at the initial part of deformation was rather poor and in this investigation a new set-up of parameters was used to describe the microstructural evolution. There is still some problems describing deformation at small strains, but the results are improved. The new set-up is used for all the data in this investigation. Using the same interaction energy for all temperatures as the value evaluated for normal grain growth at 1 200°C gives too week effect on recrystallization at lower temperature. Suehiro et al.27) suggested a linear temperature dependence on the interaction energy which was employed here. The interaction energy was set for recrystallization at 1 000°C, see Fig. 9, which together with the value for 1 200°C gave \( \Delta G_{1200} = 12 \cdot T - 30676 \). Suehiro et al. also found a dependence of the solute concentration. No apparent trend for that was seen in the calculations and it was not used here.

Figure 9 compares the calculated level of the flow stress with the measured at 1 000°C. This figure implies that the recrystallization or recovery should be more impeded at about 100 s than the model predicts. The experimental higher flow stress could depend on precipitation of AlN, MnS and/or perhaps Nb(C, N) which would further slow down both recovery and recrystallization. No precipitation is calculated here but it seems to be needed to accurately describe the experiments at 1 000°C.

The solute drag effect on recovery is described by Eq. (3) where the adjustment parameter \( k_m \) determines the extent of retardation. The \( k_m \) parameter was adjusted to get a good fit with measured flow stresses for the compression test and was set to \( 3 \cdot 10^{10} \). The result is shown in Fig. 10 to Fig. 12 where the calculated and measured stress-strain curves are illustrated. As shown in the figures, the model reproduces the experimental data quite well except for the oscillations in the curves. The oscillations in the calculated flow stress curve was discussed in an earlier investigation11) and depends on the fact that in the model, when dynamic recrystallization occurs, the material is divided into two parts, one recrystallized and one deformed. At 1 000°C and strain rate of 0.1 s\(^{-1}\), the level of the calculated curve does not really reach the measured one. It is unlikely that any precipitation has occurred and raising the value of \( k_m \) would give too high flow stresses at higher strain rates. Hence, the model can not quite describe the dislocation density evolution for this case.

![Fig. 9. Calculated (continuous line) and measured (symbols) flow stress for the Nb1 steel during stress relaxation at temperature 1000–1200°C after a pre-strain of 0.2 at a strain rate of 1.0 s\(^{-1}\).](image1)

![Fig. 10. Calculated (continuous line) and measured (broken line) flow-stress curves at strain rate 0.1 s\(^{-1}\) and various temperatures for the (a) Nb1 steel and (b) Nb2 steel.](image2)

![Fig. 11. Calculated (continuous line) and measured (broken line) flow-stress curves at strain rate 1 s\(^{-1}\) and various temperatures for the (a) Nb1 steel and (b) Nb2 steel.](image3)
A comparison between the two Nb steels and the C–Mn steel showing the effect of Nb on recovery and recrystallization is given in Fig. 13. It can be seen that the model reproduces the experimental values rather well. The calculated flow stress curves reach the level of the measured values. There is also a clear shift in the start and amount of recrystallization. The arrows indicate the start of recrystallization according to the model.

Calculations of relaxation at 1 200°C are shown in Fig. 14. The calculations are done for a pre-strain of 0.2 at strain rate 0.1 s⁻¹ and 1.0 s⁻¹. For strain rate 0.1 s⁻¹, dynamic recrystallization occurs, hence metadynamic recrystallization occurs. In the experimental results and also the calculated, recrystallization is slower in the Nb2 steel compared to the Nb1 steel. For strain rate 1.0 s⁻¹, no recrystallization occurs during deformation. Here, the difference in the measured relaxation curves between the Nb1 and Nb2 steels is not recognized. The model predicts slower recrystallization for the higher alloyed steel and the recrystallization is therefore slightly slower than the measured. Except for this slight difference, the model gives a rather good description of recovery and recrystallization, by reproducing the measured values rather well.

In Fig. 15, calculations of relaxation at 1 100°C are shown. The calculations are done for a pre-strain of 0.2 at a strain rate of 0.1 s⁻¹ and for a pre-strain of 0.4 at a strain rate of 1.0 s⁻¹. For the test at strain rate 0.1 s⁻¹, the strain is below the critical strain for dynamic recrystallization, hence static recrystallization occurs during relaxation. As shown in figure (a), the recrystallization for both steels at low strain rate is well described by the model. For the test at strain rate 1.0 s⁻¹, figure (b), dynamic recrystallization occurs for the Nb1, but not for the Nb2 steel. The best fit is achieved for the Nb1 steel. Overall, the model reproduces the measured values rather well.

In Fig. 16, calculation results for relaxation at 1 000°C after a pre-strain of 0.4 and strain rate 0.1 s⁻¹ are shown. For the Nb1 steel metadynamic recrystallization occurs and the model does not quite predict the fast recrystallization. For the Nb2 steel static recrystallization occurs and the model predicts the first part of the relaxation well but at the end either the recrystallization or the recovery is too strong. In the Nb1 steel there could be some precipitation at longer times that would retard recovery and recrystallization further and in the Nb2 steel, strain induced precipitation probably occurs rather early at 1 000°C.

To investigate further, tests at 1 050°C was conducted for the Nb2 steel. Fig. 17 shows calculated and measured relaxation tests for a pre-strain of 0.2, strain rate 0.1 s⁻¹ and a pre-strain of 0.4, strain rate 1.0 s⁻¹. Also here, the recovery and/or recrystallization are too strong at longer times.

Double compression tests at 1 050°C reveal that recrystallization is nearly complete after 200 s, the softening is about 90% using the 0.2% offset method. The model predicts 100% recrystallization after ~190 s. This means that the recrystallization should be slightly further reduced in the model. Fig. 18 shows the calculated and measured flow stress for the double compression test with the first strain of 0.4 and the second strain of 0.2. The strain applied before the relaxation test and for the first deformation of the double compression test was 0.4 and in Fig. 18(a) the measured (Exp. 1, Exp. 2) and calculated (Calc. 1, Calc. 2) values for the two tests are shown. The experiment shows some discrepancies in the flow-stress values while the model predicts almost the same flow-stress for both cases.

Although the model cannot exactly reproduce the experimental values of the relaxation as shown in Fig. 18(b), the second compression shows good fit, see Fig. 18(c). Hence, despite the lower flow stress compared to the measured at relaxation, the model well describes the mechanical properties for the following deformation. Comparison of the results of the calculated flow stress during the time between the deformations reveals that holding without applied strain gives lower values than for the relaxation test. The small applied strain during relaxation increases the calculated flow stress, but the level is still too low compared to measured data.
Fig. 14. Calculated (continuous line) and measured (symbols) flow stress for the two Nb steels during stress relaxation at temperature 1200°C after a pre-strain of 0.2 at a strain rate of (a) 0.1 s\(^{-1}\) and (b) 1.0 s\(^{-1}\).

Fig. 15. Calculated (continuous line) and measured (symbols) flow stress for the two steels during relaxation at temperature 1100°C after (a) a pre-strain of 0.2 at a strain rate of 0.1 s\(^{-1}\) and (b) a pre-strain of 0.4 at a strain rate of 1.0 s\(^{-1}\).

Fig. 16. Calculated (continuous line) and measured (symbols) flow stress for the two Nb steels during stress relaxation at temperature 1000°C after a pre-strain of 0.4 at a strain rate of 0.1 s\(^{-1}\).

Fig. 17. Calculated (continuous line) and measured (symbols) flow stress for the Nb2 steel during relaxation at temperature 1050°C after a pre-strain of 0.2 at a strain rate of 0.1 s\(^{-1}\) and a pre-strain of 0.4 at a strain rate of 1.0 s\(^{-1}\).

Fig. 18. Calculated (continuous/broken line) and measured (symbols) flow stress for the Nb2 steel for a double compression test at temperature 1050°C and strain rate of 1.0 s\(^{-1}\).
5. Conclusions

The effect of solute drag on recrystallization kinetics at temperatures above the equilibrium temperature, i.e., 1100°C, is evident when comparing the peak strain between Nb steels and a plain C–Mn steel. Also at lower temperature the retardation of recrystallization seems to be due to solute drag. When precipitation is likely to occur in the steel, the solute drag is not sufficient to reduce recovery and recrystallization.

The model to calculate flow stress takes into account the effect of Nb in solution on recovery. The calculated flow stress for the compression test showed reasonable agreement with experimental data at strain rates 0.1–10 s\(^{-1}\) and temperatures ranging from 1000–1200°C. At the lower strain rate, the model does not really reach the level of flow stress at 1000°C but at higher strain rates, the model shows good fit with measured data.

The results of calculated relaxation curves show good agreement with experimental results at temperature 1100°C and 1200°C. At the end of the relaxation tests at 1000°C and 1050°C, the recovery and recrystallization in the model are too strong.

Only by adding two models to the original model, one describing the retarding effect of solutes on recovery and one describing that on recrystallization and grain growth, the model can describe the flow stress during compression and relaxation for Nb microalloyed steel when precipitation does not occur. To be able to model the recrystallization kinetics at low temperatures in microalloyed steels a description of the precipitation kinetics is necessary.

Acknowledgement

The author expresses her gratitude to Prof. L. P. Karjalainen and Dr. M. C. Somani at Oulu University for conducting the experiments, measuring austenite grain sizes and analyzing the relaxation results. A special thanks to Prof. G. Engberg at Dalarna University who has developed the model and has given lots of useful inputs to the work done. Thanks to U. Borggren at SSAB Strip Products who helped with the Thermo-Calc calculations. Thanks for the valuable inputs by Prof. J. Ägren at KTH. Thanks also to all the colleagues at SSAB Strip Products and Dalarna University for general support and stimulating discussions. The financial support from the Swedish Knowledge Foundation and the Swedish Steel Producers’ Association (Jernkontoret) is acknowledged.

REFERENCES

1) L. J. Cuddy: Proc. of Thermomechanical Processing of Microalloyed Austenite, The Metallurgical Society/AIME, Pittsburgh, Pennsylvania, USA, (1982), 129.
2) I. Weiss and J. J. Jonas: Metall. Trans. A, 10A (1979), 831.
3) S. Yamamoto, C. Ouchi and T. Osuka: Proc. of Thermomechanical Processing of Microalloyed Austenite, The Metallurgical Society/AIME, Pittsburgh, Pennsylvania, USA, (1982), 613.
4) J. W. Cahn: Acta Metall., 10 (1962), 789.
5) M. Hillert and B. Sundman: Acta Metall., 24 (1976), 731.
6) G. Engberg: Recovery in a Titanium Stabilized 15%Cr–15%Ni Austenite Stainless Steel, R. Inst. of Technol., Stockholm, Stockholm, TRITA_MAC-98, (1976).
7) M. G. Akben, B. Bacroix and J. J. Jonas: Acta Metall., 31 (1983), 161.
8) S. F. Medina and J. E. Mancilla: ISIJ Int., 36 (1996), 1063.
9) M. Suehiro and T. Senuma: Proc. of Int. Conf. on Thermomechanical Processing of Steels and Other Materials (Thermec 97), Minerals, Metals and Materials Society/AIME, (1997), 187.
10) H. S. Zuroh, Y. Brechet and G. Purdy: Acta Mater., 49 (2001), 4183.
11) G. Engberg and L. Lissel: Steel Res. Int., 79 (2008), 47.
12) L. Lissel, G. Engberg and U. Borggren: Proc. of 3rd Int. Conf. on TMP Associazione Italiana di Metallurgica/AIM, Padua, Italy, (2008).
13) J. O. Andersson, T. Helander, L. Hoglund, P. Shi and B. Sundman: Calphad, 26 (2002), 273.
14) L. P. Karjalainen and J. Perttula: ISIJ Int., 36 (1996), 729.
15) G. Engberg: Proc. of Stål 2004, Jernkontoret, Borlänge, Sweden, (2004), 138.
16) T. Siwecki and G. Engberg: Proc. of Thermo-Mechanical Processing in Theory, Modelling & Practice [TMP] exp 2, Swedish Society for Materials Technology, Stockholm; Sweden, (1997), 121.
17) X. Wang, T. Siwecki and G. Engberg: Proc. of Int. Conf. on Processing & Manufacturing of Advanced Materials (Thermec 2003 (Part 5)), Madrid, Spain, Trans Tech Publications Ltd., Materials Science Forum Vol. 426–432 (2003), 3801.
18) Y. Bergström: The Plastic Deformation of Metals—a Dislocation Model and Its Applicability, Div. of Physical Metallurgy, KTH, Stockholm, Stockholm, (1983).
19) H. J. Frost and M. F. Ashby: Deformation-Mechanism Maps: The Plasticity and Creep of Metals and Ceramics, Pergamon Press, Oxford, (1982).
20) J. Fridberg, L.-E. Törndahl and M. Hillert: Jernkontovets Ann., 153 (1969), 263.
21) H. Oikawa: Review on Lattice Diffusion of Substitutional Impurities in Iron. A Summary Report, The Technology Reports of the Tohoku University, Vol. 47, (1982).
22) F. J. Humphreys and M. Hatherly: Recrystallization and Related Annealing Phenomena, Elsevier Science Ltd., The Boulevard, Langford Lane, Kidlington, Oxford OX5 1GB, UK, (2004), 96.
23) F. J. Humphreys and M. Hatherly: Recrystallization and Related Annealing Phenomena, Elsevier Science Ltd., The Boulevard, Langford Lane, Kidlington, Oxford OX5 1GB, UK, (2004), 114.
24) T. Hayashida, S. Sanagi and T. Kawano: Proc. of 33rd Mechanical Working and Steel Proc. Conf. Proc., The Iron and Steel Society, Warrendale, PA, USA, (1992), 79.
25) H. Avdusinovic and A. Gigovic: Metalurgija, 44 (2005), 151.
26) L. P. Karjalainen, J. Perttula, Y. Xu and J. Niu: Proc. of Physical Simulation of Casting, Hot Rolling and Welding, National Research Institute for Metals, Tsukuba, Japan, (1997), 231.
27) M. Suehiro, Z. K. Liu and J. Ägren: Acta Mater., 44 (1996), 4241.