MODELING PRECIPITATION AND ITS EFFECT ON RECRYSTALLIZATION DURING HOT STRIP ROLLING OF NIOBIUM STEELS

Linda Lissel – SSAB Tunnplåt/Dalarna University, Borlänge, Sweden
Göran Engberg – Dalarna University, Borlänge, Sweden
Ulrika Borggren – SSAB Tunnplåt, Borlänge, Sweden

ABSTRACT

Using a physically based model, the microstructural evolution of Nb microalloyed steels during rolling in SSAB Tunnplåt’s hot strip mill was modeled. The model describes the evolution of dislocation density, the creation and diffusion of vacancies, dynamic and static recovery through climb and glide, subgrain formation and growth, dynamic and static recrystallization and grain growth. Also, the model describes the dissolution and precipitation of particles. The impeding effect on grain growth and recrystallization due to solute drag and particles is accounted for. During hot strip rolling of Nb steels, Nb in solid solution retards recrystallization due to solute drag and at lower temperatures strain-induced precipitation of Nb(C,N) may occur which effectively retard recrystallization. The flow stress behavior during hot rolling was calculated where the mean flow stress values were calculated using both the model and measured mill data. The model showed that solute drag has an essential effect on recrystallization during hot rolling of Nb steels.

KEYWORDS

Modeling, steel, niobium, precipitation, hot rolling, recrystallization

INTRODUCTION

In order to get the desired product after hot strip rolling it is important to be able to predict the required roll forces. Hence, knowledge of the microstructural evolution during rolling is essential. To get the desired final mechanical properties such as a good combination of strength, fracture toughness and weldability, thermomechanical processing (TMP) is utilized [1]. By adding microalloying elements, such as Nb, V or Ti, the microstructural evolution can be controlled [2]. Microalloying elements together with TMP improve the final mechanical properties where obtaining optimum ferrite refinement is significant. This is accomplished by “maximize the area of austenite grain boundary per unit volume at the onset of phase transformation” [3]. For conventional controlled rolling, the final passes of rolling are conducted below the temperature where recrystallization occurs, i.e. below the no-recrystallization temperature, \( T_{nr} \). This results in unrecrystallized, pancaked austenite grains providing numerous nucleation sites for the ferrite and thus very fine ferrite grains [4]. Recrystallization controlled rolling on the other hand means rolling above this temperature, which results in partially or fully recrystallized grains providing fine austenite grains that give fine ferrite grains. Nb gives the most effective raise of the no-recrystallization temperature [5], therefore, in this investigation, hot rolling of Nb steels is studied. Nb impedes the movement of grain and subgrain boundaries causing a retardation of recrystallization. The retarding effect on recrystallization depends both on Nb in solid solution and on precipitated Nb(C,N) [6].
Conventional controlled rolling of Nb steels is conducted by first soaking at high temperatures, dissolving the Nb(C,N) particles. The roughing deformation is then carried out at high temperature, where recrystallization is fast and Nb is kept in solution. Finishing passes at lower temperatures then succeeds the roughing passes. At the lower temperature Nb(C,N) may start precipitating. In this investigation, the precipitation of Nb(C,N) and its effect on recrystallization during hot strip rolling is studied using a newly developed model.

1. MODEL FOR DESCRIBING THE MICROSTRUCTURAL EVOLUTION

In this work, a model describing the interactions of precipitation, recovery and recrystallization was used. The model is available in a toolbox for process simulation programmed in MATLAB® [7]. The model describes the microstructural evolution during hot deformation and is based on the work by Siwecki et al. [3] and Wang et al. [8]. The precipitation model has been shown to give reasonable agreement with measured data in a work by Borggren et al. [9]. The recrystallization model has been shown to give a good description of the behavior of CMn steels in a work by Engberg et al. [10]. It also showed satisfying results for Nb steels for both experiments and hot rolling [11] using the Dutta-Sellars model [12] for calculating the start of precipitation. To keep the calculation times fairly short, only average values are considered in the simulations.

Work hardening and recovery

The relationship between dislocation density and flow stress is described by

$$\sigma = \sigma_0 + M \cdot \alpha \cdot G \cdot b \cdot \sqrt{\rho}$$

where the first term on the right side, $\sigma_0$, consists mainly of the strengthening contribution due to second-phase particles. $M$ is the Taylor factor which is dependent on the deformation due to the development of a deformation texture [3], $\alpha$ (≈0.15) is a proportionality constant, $G$ is the temperature dependent shear modulus, $b$ is the Burgers vector and $\rho$ is the dislocation density. The flow stress is calculated for the deformed and the recrystallized part, $\sigma_{\text{def}}$ and $\sigma_{\text{rec}}$. The total flow stress for the material is calculated using a law of mixtures

$$\sigma = X_{\text{rec}} \cdot \sigma_{\text{rec}} + (1 - X_{\text{rec}}) \cdot \sigma_{\text{def}}$$

where $X_{\text{rec}}$ is the fraction recrystallized.

According to Dutta et al. [13], carbides have high elastic modulus and the particles should therefore not be sheared. Hence, Orowan strengthening [14] applies for carbides. There could also be some contribution from other particles, but this is assumed to be small and is not accounted for in the model. The increase in flow-stress due to Orowan strengthening by incoherent hard undeformable second-phase particles is given by

$$\sigma_{\text{prec}} = 0.8 \cdot M \cdot \frac{G \cdot b}{L_{\text{prec}}}$$

where $L_{\text{prec}}$ is the particle spacing. If $f$ is the volume fraction precipitates with radius $r$, the number of particles per unit volume $N_v$ (assuming spherical precipitates) is given by $3f/4\pi r^3$. With the number of particles intersecting unit area of the boundary given by $N_s = 2rN_v$, i.e. particles within a distance $r$ on either side of the boundary will intersect it, the particle spacing is given by
During hot deformation, the change in strain results in an increase of dislocation density providing work hardening, but there is also a substantial amount of recovery. In the model, recovery is described by both glide and climb. Recovery through climb is described by the diffusion of vacancies [3] and glide is proportional to the strain rate in accordance with the work by Bergström [15]. The evolution of the dislocation density with time, \( t \), is accordingly given by

\[
\frac{d\rho}{dt} = \frac{M}{bL} \frac{de}{dt} - M_m c_v \rho^2 \left[ 1 - \frac{3f}{4r\sqrt{\rho}} \right] - \Omega \rho \frac{de}{dt}
\]  

where the first term on the right side describes dislocation generation and the second and third terms describe recovery through climb and glide, respectively. In the first term, \( L \) is the dislocation mean free distance and \( \varepsilon \) is the plastic strain. In the second term, \( M_m \) is the dislocation mobility for recovery and \( c_v \) is the vacancy concentration. In the third term, \( \Omega \) is a constant (~10) [15]. Apart from the retardation on recovery due to precipitates, the recovery is also retarded by solute drag described by a reduction of the mobility of dislocations [16]

\[
M_m = \frac{M_{m0}}{1 + \sum \frac{BDx_{am}}{D_m T^2}}
\]  

where \( M_{m0} \) is the dislocation mobility without solute drag, \( B \) is a constant (~1e10), \( D \) is the diffusion for the element, i.e. Nb, \( x_{am} \) is the solute concentration in the matrix, \( T \) is the temperature, \( D_m \) is the migration part of the self diffusion coefficient of the matrix.

Recrystallization and grain growth

During deformation, when more dislocations are generated, the subgrain size, \( R_{sub} \), is reduced. In the model it is assumed that due to the substantial recovery during hot deformation the dislocations are mainly distributed at subgrain boundaries. Assuming the subgrain size, \( R_{sub} \), to be proportional to the average dislocation spacing, \( \rho^{-1/2} \), the evolution of the subgrain size is given by

\[
\frac{dR_{sub}}{dt} = -\frac{1}{2} \frac{R_{sub}^2}{\kappa_1 \rho} \frac{d\rho}{dt}
\]  

Recrystallization is driven by the stored energy of deformation and in the model, subgrains are the nuclei for recrystallizing grains. Recrystallization starts when the conditions of a critical subgrain size and configuration are fulfilled. The driving force, \( F_{rec} \), for recrystallization is dependent on the grain boundary energy of the recrystallized grains, \( \gamma_{gr} \), with radius \( R_{rec} \), the misorientation of subgrains, \( \gamma_{sub} \), with radius \( R_{sub} \), and the dislocation density and is described by

\[
F_{rec} = -\frac{\gamma_{gr}}{R_{rec}} + \frac{\gamma_{sub}}{R_{sub}} + c_d \cdot G \cdot b^2 \cdot (\rho_{def} - \rho_{rec}) - P_Z - P_S
\]  

where \( c_d \) is a constant (~0.5), \( \rho_{def} \) and \( \rho_{rec} \) are the dislocation densities in the deformed and recrystallized material, respectively, \( P_Z \) is the retarding force due to second-phase particles, i.e.
Zener pinning, and \( P_S \) is the retarding force due to solute drag. The growth of recrystallizing grains is described by

\[
\frac{dR_{gb}^3}{dt} = M_{gb}(c_v) \cdot \frac{F_{rec}^3}{3}
\]

where \( M_{gb} \) is the mobility of the boundary, which is dependent on the vacancy concentration. The fraction recrystallized in the material is the number of recrystallized grains, \( N_{rec} \), times the mean volume of the recrystallized grain. Consequently, the rate of recrystallization is described by

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

where \( c_i \) is a constant (= \( 4\pi/3 \) for spherical grains). For normal grain growth, the driving force in equation (9) is given by \( \gamma_{gb}/R_{gb} - P_Z - P_S \), where \( R_{gb} \) is the grain size and \( \gamma_{gb} \) is the grain boundary energy.

For randomly distributed particles, the Zener pinning is usually described by [17]

\[
P_Z = N_S F_S = \frac{3\gamma_{gb} f}{2r}
\]

where \( F_S \) is the maximum restraining effect on the grain boundary (= \( \pi r \gamma_{gb} \)). When the particles are instead distributed at the grain boundaries, the retarding force is expressed with the number of particles per unit area of boundary \( N_\delta = \delta N_v / S_v \), where \( S_v \) is the grain boundary area per unit volume \( (3/2R_{gb}) \) and \( \delta \) is a factor depending on the position of the particles in the boundaries, here used as a fitting parameter. Consequently, particles on grain boundaries provide a higher retarding force given by

\[
P_Z = N_\delta F_S = \frac{\delta \gamma_{gb} R_{gb}}{2r} \frac{f}{r^2}
\]

which is here used for retarding the growth of recrystallizing grains.

The retarding force due to solute drag is only considered for Nb. In the model, the retarding force due to solute drag is described by the simplified model by Cahn [18]

\[
P_S = \alpha_S \chi^{am} \frac{dR_{gb}}{dt}
\]

where the parameter \( \alpha_S \) is given by

\[
\alpha_S = \frac{wN_v (kT)^2}{ED} \left[ \sinh \left( \frac{E}{kT} \right) - \frac{E}{kT} \right]
\]

where \( w \) is the grain boundary width, \( N_v \) is the number of atoms per unit volume, \( E \) is the interaction energy between Nb and the grain boundary and \( k \) is Boltzmann constant. In a work by Zurob et al. [19] the values was set to \( w \sim 1 \) nm and \( E = 25000 \) J/mol. In this investigation the values of \( w = 0.5 \) nm and \( E = 30000 \) J/mol gave good agreement with experiments, see Fig. 1. In the figure,
calculations with the model is shown for a steel with composition 0.065C, 0.008N and 0.034Nb given in wt%. The figure shows compression and stress relaxation tests for temperatures 1000, 1100 and 1200°C, a pre-strain of 0.4 and a strain rate of 1 s\(^{-1}\). The precipitation model, described below, predicted some precipitation of NbC at 1000°C at longer times.

![Graphs of stress-strain and stress relaxation tests](image)

**Precipitation model**

In the Fe-M-X ternary system considered here, the MX-type precipitates, where M is one of the alloying elements Mn, Al and Nb and X is one of the elements C, N and S, the diffusivity of element X is several orders of magnitude higher than that of element M, see Table 1. Therefore, it is assumed that the element with the lowest diffusion rate, element M, is the rate-controlling element. It is also assumed that the concentration gradient of element X is negligible.

**Table 1. Diffusion coefficients used in the calculations.**

<table>
<thead>
<tr>
<th></th>
<th>Fe</th>
<th>Mn</th>
<th>Al</th>
<th>Nb</th>
<th>S</th>
<th>N</th>
<th>C</th>
</tr>
</thead>
<tbody>
<tr>
<td>D(_0) [m(^2)/s]</td>
<td>7·10(^{-5})</td>
<td>1.78·10(^{-5})</td>
<td>2.1·10(^{-3})</td>
<td>5.6·10(^{-5})</td>
<td>7.52·10(^{-4})</td>
<td>3.72·10(^{-5})</td>
<td>3.72·10(^{-5})</td>
</tr>
<tr>
<td>Q [J/mol]</td>
<td>286000</td>
<td>264000</td>
<td>286000</td>
<td>286000</td>
<td>236400</td>
<td>142360</td>
<td>148220</td>
</tr>
<tr>
<td>Reference</td>
<td>[20]</td>
<td>[21]</td>
<td>[20]</td>
<td>[20]</td>
<td>[21]</td>
<td>[22]</td>
<td>[22]</td>
</tr>
</tbody>
</table>

In the model, solubility products are used to describe the solubility of the precipitates

\[
\log[M][X] = A - \frac{B}{T}
\]  

(15)

where [M] and [X] are the wt% of the elements participating in the precipitate. The solubility products used are shown in Table 2.

**Table 2. Solubility product data used in the calculations.**

<table>
<thead>
<tr>
<th></th>
<th>MnS</th>
<th>AlN</th>
<th>NbC</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>2.93</td>
<td>1.03</td>
<td>2.26</td>
</tr>
<tr>
<td>B</td>
<td>9020</td>
<td>6770</td>
<td>6770</td>
</tr>
<tr>
<td>Reference</td>
<td>[23]</td>
<td>[24]</td>
<td>[25]</td>
</tr>
</tbody>
</table>
Assuming no concentration gradient of element X gives a deviation from complete local equilibrium. Therefore, for the dissolution and precipitation of a secondary phase $\beta$ in a matrix phase $\alpha$, a para-equilibrium with concentration $x^{\alpha\beta}_{\text{para}}$ is considered. Schematic diagrams of the solute concentration around a dissolving and growing secondary phase $\beta$ (MX-type precipitate) in a matrix phase $\alpha$ are shown in Fig. 2 and Fig. 3, respectively.

![Fig. 2. Schematic diagram of the concentration profile of a dissolving precipitate and the isothermal phase diagram for Fe-M-X ternary system. The details in the figure are explained in the text.](image)

In the figures, $r$ is the particle radius, $x^\beta$ is the concentration of the growing $\beta$-phase, $x^{\alpha\beta}$ is the equilibrium concentration assuming complete local equilibrium at the matrix-particle interface.
whereas $x_{\text{para}}^{a\beta}$ is the concentration at the matrix-particle interface assuming no concentration gradient of element X. The term $x^a$ is the matrix concentration, $x_{\text{nom}}$ is the nominal composition, $x^a_0$ is the concentration of the $\alpha$-phase at a distance larger than or equal to $R_{\text{diff}}$, with $R_{\text{diff}}$ being the diffusion distance from the particle, i.e. the distance where the concentration gradient levels out. The distance $R_{\text{max}}$ is the maximum diffusion distance, i.e. half the distance between particles. In the figures, the tie-line, assuming complete local equilibrium is the dashed line in the phase diagram. The tie-line drawn with a continuous line is for assuming concentration gradients only for element M.

The results of studies of the precipitation kinetics of microalloying elements and their effect on the microstructure show that prior deformation enhances the precipitation kinetics significantly [12]. Precipitation occurs preferably on defects, such as grains and dislocations, i.e. subgrain boundaries assuming the dislocations lie in subgrain boundaries. For the process data considered here, homogenous nucleation is assumed negligible and nucleation is assumed to occur heterogeneously on grain boundaries and subgrain boundaries.

The precipitation model is based on classical nucleation theory [26]. The nucleation rate per unit volume is proportional to the grain boundary area, $S_v$, and is described by

$$\frac{dn}{dt} = k_1 \cdot x_{M}^a \cdot D^\beta \cdot e^{\frac{k_2 \cdot \Delta G}{kT}}$$  \hspace{1cm} (16)

where $n$ is the number of precipitates per unit volume, $k_1$ and $k_2$ are a proportionality constant, $\Delta G$ is the activation energy, the grain boundary area $S_v$ is $X_{\text{rec}} / R_{\text{rec}} + (1 - X_{\text{rec}}) / R_{\text{def}}$ for grain boundary nucleation and for subgrain boundary nucleation $S_v = X_{\text{rec}} / R_{\text{sub,rec}} + (1 - X_{\text{rec}}) / R_{\text{sub,def}}$.

The parameters $k_1$ and $k_2$ was tuned for strain-induced precipitation using literature data with experiments at different temperatures [27]. Weiss and Jonas [27] studied the interaction between recrystallization and precipitation and compared their results with other investigations. They came to the conclusion that the technique used to follow the process has significant effect on the precipitation rates determined experimentally. In this work, the effect of the precipitates on recrystallization is of interest and the parameters were set to be able to reproduce experimental data of stress-relaxation tests. Good agreement was obtained between calculated and measured values as is shown in Fig. 4, where the calculation of a compression test followed by stress relaxation at temperature 900°C for the same steel as in Fig. 1 is presented. For grain boundary nucleation $k_1$ and $k_2$ was set to $2 \cdot 10^{33}$ and 0.1, respectively, for subgrain boundary nucleation $k_1$ and $k_2$ was set to $2 \cdot 10^{33}$ and 0.24, respectively.

![Fig. 4. Calculated (dashed) and experimental (solid) values for (a) a stress-strain curve and (b) a stress relaxation curve at 900°C ($\varepsilon = 0.4$, $de/dt = 1$ s$^{-1}$).](#)
According to Dutta et al. [13] the core energy of a dislocation line gives a small contribution to the activation energy for nucleation but it is not considered here. Consequently, in the present model, the contributions to the activation energy is given by the free energy from the chemical supersaturation \( \Delta G_{\text{chem}} = 4/3 \pi R^3 \Delta G_v \) and the creation of incoherent interfaces \( \Delta G_{\text{int}} = 4\pi R^2 \gamma \) which together with the critical radius for nucleation expressed as \( R_c = 2\gamma/\Delta G_v \) gives a total activation energy of

\[
\Delta G = \frac{16}{3} \pi \frac{\gamma^3}{(\Delta G_v)^2} \quad (17)
\]

Where \( \gamma \) is the interfacial energy of the precipitate and \( \Delta G_v \) is the driving force for precipitation, given by [28]

\[
\Delta G_v = -\frac{R_{\text{gas}} T}{V_m^\beta} \ln \left( \frac{x_m^{\alpha m} y_m^{\alpha m}}{x_m^{\beta \alpha} y_m^{\beta \alpha}} \right) \quad (18)
\]

where \( R_{\text{gas}} \) is the gas constant, \( V_m^\beta \) is the molar volume of the \( \beta \)-phase, \( x_m^{\alpha m} \) and \( y_m^{\alpha m} \) are the matrix concentration in mol fraction of element \( M \) and \( X \), respectively, and \( x_m^{\beta \alpha} \) and \( y_m^{\beta \alpha} \) are the para-equilibrium concentrations in mol fraction of element \( M \) and \( X \), respectively.

For the solution of Fick’s law, a quasi-stationary approximation assuming spherical growth is used [29]

\[
\frac{dr}{dt} = D \frac{\Theta}{r(1-r/R_{\text{diff}})} \quad (19)
\]

where \( \Theta \) is the supersaturation

\[
\Theta = \frac{V_m^\alpha}{V_m^\beta} \left( \frac{x_m^{a0} - x_m^{\alpha \beta}}{x_m^{\beta \alpha} - x_m^{\alpha \beta}} \right) \quad (20)
\]

where \( V_m^\alpha \) is the molar volume of the \( \alpha \)-phase, \( x_m^{a0} \) is the concentration at the distance \( R_{\text{diff}} \) from the particle. The diffusion distance, \( R_{\text{diff}} \), is calculated with a mass balance

\[
\frac{4}{3} \pi \frac{3}{2} \pi \frac{x_m^{\beta \alpha}}{V_m^\alpha} + \int_r^{R_{\text{diff}}} x_m^\alpha(y)4\pi y^2 dy + \frac{4}{3} \pi \left( R_{\text{max}}^3 - R_{\text{diff}}^3 \right) x_m^{a0} = \frac{4}{3} \pi R_{\text{max}}^3 x_{\text{nom}} \quad (21)
\]

The gradient of \( x_m^\alpha \) as a function of the distance \( y \) is given by (for spherical particles)

\[
\frac{1}{y} - \frac{1}{r} = \frac{x_m^\alpha(y) - x_m^{\alpha \beta}}{x_m^{a0} - x_m^{\alpha \beta}} \quad (22)
\]

where \( x_m^\alpha(y) \) is the concentration of the \( \alpha \)-phase at a distance \( y \) from the particle centre. When \( R_{\text{diff}} = R_{\text{max}} \), soft impingement occurs and equation (21) is used to determine the value of \( x_m^{a0} \).

When the equilibrium fraction is reached, Ostwald ripening occurs, but in these simulations, no Ostwald ripening is accounted for due to the fact that the equilibrium is never reached.
2. MATERIAL AND EXPERIMENTAL PROCEDURE

For the purpose of investigating the effect of precipitation on recrystallization during hot strip rolling of Nb steels, the microstructural evolution during hot rolling was calculated for steel grades with different Nb content. Three commercial steel grades rolled in SSAB’s hot strip mill (Domex 240, Domex 420, Domex 490) were selected. The chemical composition is presented in Table 3. The final thickness of the strips after hot rolling was 8 mm.

Table 3. Chemical composition of the steels investigated (wt%).

<table>
<thead>
<tr>
<th></th>
<th>Mn</th>
<th>C</th>
<th>Si</th>
<th>P</th>
<th>S</th>
<th>N</th>
<th>Al</th>
<th>Nb</th>
</tr>
</thead>
<tbody>
<tr>
<td>Domex 240</td>
<td>0.065</td>
<td>0.064</td>
<td>0.007</td>
<td>0.011</td>
<td>0.006</td>
<td>0.006</td>
<td>0.041</td>
<td>-</td>
</tr>
<tr>
<td>Domex 420</td>
<td>1.100</td>
<td>0.064</td>
<td>0.010</td>
<td>0.009</td>
<td>0.003</td>
<td>0.005</td>
<td>0.046</td>
<td>0.036</td>
</tr>
<tr>
<td>Domex 490</td>
<td>1.360</td>
<td>0.064</td>
<td>0.040</td>
<td>0.009</td>
<td>0.002</td>
<td>0.005</td>
<td>0.048</td>
<td>0.063</td>
</tr>
</tbody>
</table>

Process data for the three steels was collected from the mill logs of SSAB’s hot strip mill. The mill data consisted of the soaking temperature in the reheating furnace, the reduction and rolling speed in each pass and also the temperature. The temperature of the workpiece is only measured at a few places in the mill and only at the surface. Hence, the temperature of the material had to be calculated. The temperature was calculated using the multi-node temperature model in HSMM [30] with a difference of about ±10°C from the measured surface temperature. The average strip temperature from the HSMM-model was used as input to the microstructure model.

The only measurements from the hot strip mill that can be used to validate the accuracy of the calculations are the measured roll forces. In order to compare the measured roll forces with the calculated flow stress, the mean flow stress of each pass was calculated as defined by

\[ MFS = \frac{1}{\varepsilon_1 - \varepsilon_0} \int_{\varepsilon_0}^{\varepsilon_1} \sigma d\varepsilon \]  

where \( \varepsilon_0 \) is the initial strain, \( \varepsilon_1 \) is the total strain. The mean flow stresses determined with the microstructure model were compared to the mean flow stresses determined from the measured roll force, calculated using a friction-hill roll-force model. As input data to the friction-hill model, measured process data, such as measured roll forces and geometry of the strip were used and also a temperature-dependent friction-coefficient.

3. RESULT AND DISCUSSION

In the calculations, the interfacial energies for precipitation of MnS, AlN and NbC were set to 1, 0.75 and 0.5, respectively. The diffusion coefficients and the solubility products used are given in Table 1 and Table 2, respectively.

In the roughing mill, all three materials recrystallizes and no precipitation of NbC occurred according to the model. Therefore, only calculations of the finishing mill are discussed. Comparison of the mean flow stresses from measured data compared to mean flow stresses calculated with the microstructural model for the three steels are shown in Fig. 5. The figure shows the mean flow
stress behavior in the six pass finishing mill. The agreement between calculated and measured values is reasonably good, implying that the calculations give accurate values for recrystallization and precipitation kinetics. For Domex 420, the measured mean flow stress values are not quite met, maybe there is some strengthening contribution missing and there should in fact be some more precipitation at lower temperatures.

Fig. 5. Comparison of the mean flow stress (MFS) from measured data to the mean flow stress calculated with the microstructural model for rolling in the finishing mill for the three investigated steels.

Domex 240 is a plain CMn steel and, as is shown in Fig. 5, the flow stress is lower then for the Nb steels. For Domex 240, recrystallization is almost complete throughout the whole process, except between the last three passes where partial recrystallization occurs. Rather small austenite grains are obtained (15 µm) which should result in fine ferrite grains upon transformation.

Calculated precipitation of NbC and fraction recrystallized during rolling in the finishing mill for Domex 420 and Domex 490 is shown in Fig. 6. In the figure, the six finishing passes are numbered F1 to F6. For Domex 420, recrystallization is completely retarded in the last passes. There is very little precipitation of NbC, hence the retardation of recrystallization is due to Nb in solution. The final austenite grain size is predicted to be ~40 µm. For Domex 490, precipitation of NbC occurs, but there is still Nb in solution and therefore an effect of solute drag. The recrystallization is completely retarded in the last passes due to precipitation and solute drag. The final austenite grain size is predicted to be ~45 µm. For both Domex 420 and Domex 490, the accumulation of strain, i.e. the pancaking of the austenite grains, should provide very fine ferrite grains after transformation.

Fig. 6. Volume fraction precipitated NbC and fraction recrystallized during rolling in the finishing mill for (a) Domex 420 and (b) Domex 490.
4. CONCLUSIONS

The mean flow stress calculated with the microstructural model is in good agreement with the measured values. Only by using mean size values for particles and a simplified solute drag model a good description of the microstructural evolution during hot rolling is achieved.

The model indicates that for a plain CMn steel, there is almost complete recrystallization throughout the process. For the two Nb steels investigated here, a noticeable amount of NbC precipitation occurs only for the higher alloyed steel. Nevertheless, in the last passes in the finishing mill, recrystallization is retarded for both the Nb steels. This means that solute drag has an essential effect on the retardation of recrystallization.

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