1. Introduction

In recent years a significant amount of attention has been paid to developing mathematical models that predict the final microstructure of a hot rolled product as a function of the chemical composition and processing conditions. The application of these models allows the optimization of deformation sequences and so obtain a good combination of mechanical properties in thermomechanically processed steels.

The metallurgical events that occur during rolling are very sensitive to the operating conditions and steel composition. In the case of Nb microalloyed steels, the influence of Nb on the metallurgical processes that occur during rolling must also be taken into account. Nb is thought to contribute to a strengthening of steel by two mechanisms. On the one hand, it retards recrystallization kinetics during hot rolling, both by the solute drag effect due to the presence of solute atoms and by the pinning effect due to strain induced precipitates, the latter usually exerting the stronger effect. This helps to retain the accumulated strain and deformed structures of austenite grains. As a result, the ferrite grain size obtained after transformation is finer compared to the one obtained from a recrystallized austenite microstructure, which leads to an improvement in the mechanical properties of the final product. On the other hand, if Nb remains in solution prior to the \( \gamma \) to \( \alpha \) transformation, the increase in hardenability, as well as in the precipitates which form during or after transformation, contribute to an increase in the strength of the steel. This means that knowledge of the kinetics of precipitation in austenite is a critical issue when optimizing deformation sequences.

A large number of researchers have studied the precipitation kinetics of Nb in austenite. However, these studies involve a number of variables, such as steel composition, deformation conditions and detection methods, which make a direct comparison between the results difficult. Recently, there has been a tendency to develop physically based models to study strain-induced precipitation. However, these models require the fit of physical unknown parameters, and their application to industrial conditions continue to be difficult. As a consequence, empirical and semi-empirical models remain a good alternative for understanding and improving industrial sequences. Nevertheless, one of their main limitations continues to be their inaccurate abilities of prediction when they are applied to steel grades other than those initially employed to determine the equations.

In this paper a previously developed microstructural model was improved on, with special attention paid to the development of a more general expression for the prediction of the onset of strain induced precipitation in Nb microalloyed steels. To that purpose the equation initially proposed by Dutta and Sellars was modified by considering new data available from literature which take into account a wide range of steel compositions and deformation conditions. Moreover, the evolution of strain induced Nb precipitation was modeled in order to determine the amount of Nb precipitated after each strain pass.
2. Model Development

2.1. Static Recrystallization

The evolution of austenite microstructure after each deformation was modeled. Only those conditions for which the effective strain is less than $\varepsilon_c$ (the critical strain for the onset of dynamic recrystallization) were considered, i.e. only classical static recrystallization kinetics.

The evolution of the recrystallized fraction with time is described by the Avrami equation:

$$X = 1 - \exp\left( - \ln 2 \left( \frac{t}{t_{0.5X}} \right)^n \right) \quad \text{(1)}$$

where $t$ is the time (s), $t_{0.5X}$ is the time to reach a 50% recrystallized fraction and $n$ is the Avrami exponent.

In a previous work, the static recrystallization kinetics and the retardation that Nb in solution exerted on the recrystallization time were studied. Equation (2) was developed to calculate the time for 50% recrystallization fraction, $t_{0.5RX}$:

$$t_{0.5RX} = 9.92 \times 10^{-11} D_0 e^{-5.6D_0^{0.56} \varepsilon\cdot\varepsilon} \cdot \exp\left( \frac{180000}{RT} \right) \times \exp\left( \frac{275000}{T} - 185 \right) \{ \text{Nb} \} \quad \text{(2)}$$

where $D_0$ represents the initial austenite grain size, $\varepsilon$ and $\dot{\varepsilon}$ are the applied strain and strain rate respectively, $T$ is the absolute deformation temperature and $\{ \text{Nb} \}$ represents the niobium concentration in solution in wt%.

Equation (2) is used in the present model to describe recrystallization kinetics.

Equation (2) was determined over a range of high temperatures, between 1100 to 1150°C, where the value of the Avrami exponent $n$ was found to be almost independent of the deformation parameters and with a mean value close to $n=1$. However, when a wider range of temperatures was considered, a dependence of $n$ on temperature has been reported for microalloyed steels. Medina et al. determined the following equation for microalloyed steels, which predicts a slight decrease in $n$ with decreasing temperature:

$$n = 28 \exp\left( \frac{-36000}{RT} \right) \quad \text{(3)}$$

Equation (3) was determined in the range of deformation conditions where strain induced precipitation did not occur. However, in a recent work it was found that when strain induced precipitation interacted with softening processes the exponent $n$ took values even lower than those given by Eq. (3). The following equation was derived in this case:

$$n = 155 \exp\left( \frac{-56485}{RT} \right) \quad \text{(4)}$$

Consequently, Eq. (3) was used for modeling recrystallization kinetics in the range of deformation passes before the onset of strain induced precipitation whereas Eq. (4) was applied afterwards. It was shown that using these relationships brought better predictions than assuming a constant value for the Avrami exponent around $n=1$ in the overall temperature range.

The equations above were established for isothermal conditions; however during rolling simulations the temperature decreases continuously. The principles of the additivity rule were used to modify the isothermal recrystallization model for use in continuous cooling conditions. The cooling curve is divided into increments of time, $\Delta t$, and the temperature drop for each step can be calculated as $\Delta T = \Delta t \cdot v$, $v$ being the cooling rate. If the accumulated recrystallized fraction at the end of an interval of temperature $T_i$ is $X_i$, the recrystallized fraction at the end of the subsequent interval of temperature $T_{i+1} < T_i$ is calculated by first determining the “equivalent time” $t_{eq}$ needed to reach the previous recrystallized fraction value of $X_i$ during the isothermal annealing at $T_{i+1}$. The real recrystallized fraction at $T_{i+1}$ is then calculated as:

$$X_{i+1}(\Delta t) = X_{i+1}(\Delta t + t_{eq}) - X_{i+1}(t_{eq}) \quad \text{(5)}$$

The recrystallized fraction is calculated and added from interval to interval. By adding the holding time for each increment, the time necessary to obtain a certain recrystallized fraction can also be determined.

The statically recrystallized grain size after each pass $d_{ex}$ is estimated by the following equation:

$$d_{ex} = 1.4 \left( \frac{D_0^{0.56}}{\varepsilon} \right) \quad \text{(6)}$$

To describe the kinetics of grain growth after recrystallization the following equation proposed by Hodgson and Gibbs for Nb bearing steels is applied:

$$D^{4.5} = D^{4.5}_{ex} + 4.1 \times 10^{23} t_{eq} \exp\left( - \frac{435000}{RT} \right) \quad \text{(7)}$$

where $t_{eq}$ is the time remaining after complete recrystallization, assuming a 95% recrystallized fraction ($t_{eq} = t_{sp} - t_{0.95})$. The continuous decrease in temperature is also taken into account through the application of the additivity rule to grain growth kinetics.

In multipass rolling, it is possible to have partial recrystallization after a pass strain, $\varepsilon_i$, if the time between passes is not long enough for complete recrystallization. This introduces a mixed microstructure before the next deformation pass, $\varepsilon_{i+1}$. To cope with this effect the “uniform softening method” was used. This method considers a single average microstructure, described by an average grain size, with an effective strain as follows:

$$\varepsilon_{eff} = \varepsilon_i + \lambda (1 - X_{i-1}) \varepsilon_{i-1} \quad \text{(8)}$$

where $\lambda$ is a constant which is taken as 0.5 for Nb steels and $X_{i-1}$ is the recrystallized fraction between passes $i$ and $i-1$. This effective strain must be taken into consideration for the calculations of recrystallization and strain-induced precipitation kinetics after each pass.

In this method, the partially recrystallized microstructure is described by an average grain size. Different expressions have been proposed in literature to calculate this grain.
size.\(^{21,22}\) In a previous study\(^{23}\) it was observed that an approach that takes the average grain size as being equal to the fully recrystallized grain size provided a good correlation with experimental results, once the recrystallized fraction had overcome a minimum value. In the present work, the same criterion was adopted for modeling the evolution of the mixed microstructure when partial recrystallization was predicted during rolling simulations:

\[
\bar{d} = d_{\text{rec}} \text{..........................(9)}
\]

2.2. Nb(C, N) Strain-induced Precipitation

2.2.1. Precipitation Constants

For microalloyed steels recrystallization kinetics is delayed when strain-induced precipitation occurs. A model to predict the start of strain induced precipitation in Nb microalloyed steels was first proposed by Dutta and Sellars.\(^6\) Based on nucleation theory these authors proposed the following equation for calculating the time taken for 5\% Nb(C, N) precipitation to occur:

\[
t_{0.05p} = A[Nb]^{-1}e^{-1Z^{-0.5}} \exp \left( \frac{270000}{RT} \right) \exp \left( \frac{B}{T^3(\ln k_s)^2} \right)
\text{..........................(10)}
\]

where \(A\) and \(B\) are constants and \(k_s\) which were calculated to fit available experimental data for Nb-bearing steels, \(Z\) is the Zener–Hollomon parameter \((Z=t\exp(Q_{\text{act}}/RT))\) with \(Q_{\text{act}}\) the activation energy for deformation, \(k_s\) represents the supersaturation ratio at the temperature \(T\), defined as the ratio of the actual amount of \([Nb][C+(12/14)N]\) in solution to the equilibrium amount which is calculated by the solubility product equation. Dutta and Sellars, using a value of \(Q_{\text{act}}=400\ kJ/mol\) and the Irvine \(k_s\) solubility product,\(^{24}\) calculated mean values of \(A=3\times10^{-6}\) and \(B=2.5\times10^{10}\) by fitting the experimental data available.

With these constants it has been reported that for some compositions and deformation ranges the predicted values do not agree with the times measured for the onset of precipitation.\(^{12,25,26}\) This limitation has been overcome by proposing new values for the constants present in Eq. (10) but with all the amounts in percentage weight. Within the scatter a trend is clearly observed for the value of \(B\) to increase as the \([Nb][C+(12/14)N]\) product increases. Nearly one order of magnitude difference is observed in the composition range analyzed. This means that assuming a constant \(B\) value for all the compositions could lead to large errors. In view of the results, the following equation was proposed (without considering the data corresponding to steels H and O in the fit as these deviated significantly from the general trend):

\[
B = 1.59 \times 10^9 \exp(459 - ([Nb][C + (12/14)N])) \text{.........(12)}
\]

In Fig. 2 the optimum values of \(A\) were calculated by intro-

![Fig. 1. Optimum value for B in Eq. (10) obtained by applying minimum squares method to the strain-induced precipitation start times reported for the steels in Table 1.](image)

Table 1. Composition of the steels (wt\%) and strain-induced precipitation detection method used to determine constants \(A\) and \(B\) in Eq. (10).

| Steel | C   | N   | Nb  | Ref. | \(t_{0.05p}\) detection method         |
|-------|-----|-----|-----|------|----------------------------------------|
| A     | 0.084| 0.061| 0.03| 9    | Electrolytic extraction and centrifuging |
| B     | 0.08 | 0.015| 0.03|      | Stress relaxation-Compression tests     |
| C     | 0.084| 0.015| 0.06|      | Stress relaxation-torsion tests         |
| D     | 0.081| 0.014| 0.12|      | Two-stage interrupted compression tests  |
| E     | 0.062| 0.0082| 0.051| 31   | Two-stage interrupted torsion tests     |
| F     | 0.19 | 0.0088| 0.015| 32   | Extraction replica                      |
| G     | 0.026| 0.001| 0.11| 26   | Microhardness                           |
| H     | 0.076| 0.0061| 0.03| 33   | Two-stage interrupted compression tests  |
| J     | 0.29 | 0.0662| 0.066| 34   | Two-stage interrupted torsion tests     |
| K     | 0.21 | 0.0058| 0.023| 35, 36|                                            |
| L     | 0.2  | 0.0056| 0.007| 17   |                                            |
| M     | 0.51 | 0.0105| 0.026| 38   |                                            |
| N     | 0.11 | 0.001| 0.031| 8    |                                            |
| O     | 0.10 | 0.001| 0.095| 8    |                                            |
| P     | 0.019| 0.0028| 0.095| 39   |                                            |
| Q     | 0.18 | 0.0054| 0.031| 40   |                                            |
Precipitation is that it cannot separate boundary precipitation.

Very interesting to note that the scatter in the estimated values of precipitation can be fitted independently, leading to the following equation:

\[ A_{\text{predicted}} = 3.70 \times 10^6 \exp(-372[Nb][C+(12/14) N]) \]

In Fig. 2, the optimum value for \( A \) in Eq. (10) obtained for the steels of Table 1 when Eq. (12) is used for the calculation of \( B \).

The main disadvantage related to the chemical extraction of precipitate is that it cannot separate boundary precipitation and matrix precipitation, and as a result, it may lead to precipitation start times shorter than in other methods of precipitation detection.

Concerning the rest of the data, it is interesting to note that the values of \( A \) calculated for steels E–M where mechanical testing methods were used, these include stress-relaxation, double compression or double torsion tests (plateaus in softening curves), tend to be lower than those estimated for steels N–Q, where extraction replica (N–P) and microhardness measurements (Q) were employed. The following equations were derived in both cases:

\[ A_{\text{chem.extract}} = 9.42 \times 10^{-7} \exp(-372 \cdot ([Nb] \cdot [C + (12/14) \cdot N])) \]

\[ A_{\text{mechanical}} = 8.74 \times 10^{-6} \exp(-372 \cdot ([Nb] \cdot [C + (12/14) \cdot N])) \]

for the data corresponding to mechanical testing detection methods and

\[ A_{\text{replica-hardness}} = 3.70 \times 10^{-5} \exp(-372 \cdot ([Nb] \cdot [C + (12/14) \cdot N])) \]

when replica or microhardness measurements were used.

It must be pointed out that the application of Eqs. (12)–(15) should be limited to the range of compositions they were derived at: nominal values of composition of %C from 0.019 to 0.51 and %Nb from 0.007 to 0.12% and values of the product \( Nb \cdot [C + (12/14) \cdot N] \) from 0.00143 to 0.0063, assuming the available microalloying content given by the solubility product of Eq. (11).

In Fig. 3 the predictions of the present precipitation model and the Dutta and Sellars model are compared to the experimental PTT diagrams of the steels A and D (electrolytic extraction), E and G (stress relaxation), K and L (interrupted torsion test) and N and P (extraction replica) from Table 1. It can be seen that the proposed equations lead to more accurate results than the Dutta and Sellars model (except steel D where similar predictions were obtained following both models). In some cases the new model reasonably predicts the experimental results (steels A, E, K and L), while in other cases, some differences remain when compared to experimental values (steels G and N) although a significant improvement is achieved.

2.2.2. Amount of Nb Precipitated

Nb precipitation, just like recrystallization, is a process that involves nucleation and growth events. As a result, the kinetics of Nb(C, N) precipitation may be also modeled by an avrami type equation:

\[ X_p = 1 - \exp \left( \frac{t}{t_{0.65p}} \right)^{n_p} \]

where \( X_p \) is the precipitated fraction for a time \( t \), \( n_p \) the Avrami exponent for precipitation and \( t_{0.65p} \) represents the start time for strain-induced precipitation given by Eq. (10). For the Avrami exponent in Eq. (16) a value of \( n_p = 0.6 \), taken from Ref. 5, was considered. This value is in the range of those reported by other authors.

For a constant temperature, the maximum precipitated volume fraction depends on the equilibrium conditions. This equilibrium precipitated volume fraction can be calculated supposing a precipitate stoichiometry and a solubility product. The solubility product of Eq. (11) was used in this...
study so as to be consistent with the above calculations for recrystallization and precipitation time. Thus, the precipitated volume fraction for a constant temperature is calculated as:

\[ f_v = f_{\text{eq}} \cdot X_p \]  \hspace{1cm} (17)

where \( f_{\text{eq}} \) is the equilibrium volume fraction at the temperature considered and \( X_p \) is the precipitated fraction calculated from Eq. (16).

As in the case of static recrystallization, Eqs. (10), (16) and (17) were obtained under isothermal conditions. In the case of multipass rolling, the continuous decrease in temperature must be taken into account. In this study, the time required for the onset of precipitation after deformation was also calculated by applying the additivity rule, using the relation:

\[ \sum t_i \Delta t f_{0.05p}(T_j) = 1 \]  \hspace{1cm} (18)

where \( t_{0.05p}(T_j) \) is the time for 5% precipitation at the time interval of temperature \( T_j \) and \( T_i \) is the deformation temperature for the \( i \)-th pass. The precipitation start time is evaluated at each interval from the deformation temperature and when the summatorial \( \Delta t / t_{0.05p}(T_j) \) reaches unity, strain-induced precipitation is considered to commence and recrystallization is assumed to have stopped. Thus, the level of recrystallized fraction reached after the \( i \)-th pass corresponds to the accumulated recrystallized fraction calculated at the interval \( j \) that fulfills the condition of Eq. (18).

Equation (18) provides the conditions for the beginning of strain induced precipitation after a given deformation pass. However, the evolution of the precipitated fraction (Eq. (16)) must be also determined, since once precipitation starts more and more Nb is taken out of solution and therefore a smaller amount of Nb is available for further precipitation and solute drag effect after the subsequent pass. Similarly with recrystallization, the additivity rule is applied to Eq. (16) for the calculation of the evolution of precipitated fraction at continuous cooling conditions. The following equation is used to calculate the precipitated fraction at the end of an interval of temperature \( T_j \):

\[ X_{p+1}(\Delta t) = X_{p+1}(\Delta t + t_{eq}) - X_{p+1}(t_{eq}) \]  \hspace{1cm} (19)

Where as in recrystallization, \( t_{eq} \) represents the “equivalent time” required to reach the precipitated fraction \( X_p \) present at the end of the previous interval \( T_j \) during the isothermal annealing at \( T_{j+1} \).

The equilibrium precipitated volume fraction \( f_{eq} \) is calculated for each interval, and so the real precipitated volume fraction \( f_p(\Delta t) = f_{eq}(\Delta t) \cdot f_{eq} \) can be obtained by using Eq. (17). The values are added from interval to interval and the total precipitated volume fraction, together with the amount of \([\text{Nb}]\) and effective carbon \([C+(12/14)N]\) concentrations that remain in solution at the end of the interpass time, are calculated. In the next pass, all summator-
als are set to zero and the calculations begin again, but introducing the new amounts of [Nb] and effective carbon [C+(12/14)N] in solution in Eqs. (2) and (10) for the calculation of recrystallization and precipitation kinetics after the new pass. In this way, the model predicts the recrystallized fraction and the amounts of Nb remaining in solution and precipitated after each pass throughout the overall deformation schedule.

2.3. Structure of the Model

In Fig. 4 the structure of the model is shown using a flow chart. The first step is the calculation of the actual amount of [Nb] and [C+(12/14)N] in solution for each reheating temperature according to the solubility product of Palmiere et al. For each pass, the model predicts three possible final situations: partial recrystallization without precipitation; grain growth after complete recrystallization and partial recrystallization due to the onset of strain induced precipitation. In the latter case, the precipitated volume fraction is calculated.

In the next pass the new initial grain size, [Nb] and [C+(12/14)N] concentrations in solution and the accumulated strain are considered for the calculations. Precipitation constants in Eq. (10) are taken to depend on Nb, C and N in solution, and are also recalculated for each pass. After the last pass, an output file including the recrystallized fraction, austenite grain size, Nb(C, N) precipitated volume fraction and dissolved [Nb] and [C+(12/14)N] amounts calculated after each pass is generated.

In summary the following improvements have been introduced into the model in comparison to the previous version:

- Dependence of the Avrami exponent for recrystallization, \( n \), with temperature and presence of precipitation.
- Dependence of constants \( A \) and \( B \) on the precipitation equation with the chemical composition (C, N and Nb).
- Determination of the remaining amount of Nb in solution as precipitation proceeds and available for subsequent precipitation sequences.

3. Experimental

For the validation of the microstructural model a series of multipass torsion tests were carried out at decreasing temperature on six Nb microalloyed steels, their compositions are given in Table 2. With the selected steels a range of carbon contents from 0.05 to 0.14% and Nb contents from 0.019 to 0.060% were considered. The torsion samples had a gauge length of 17 mm and a diameter of 7.5 mm. Specimens were reheated in an induction furnace for 15 min, using high reheating temperatures, in the range from 1350 to 1440°C, to ensure the complete dissolution of Nb precipitates.

After reheating, the specimens were deformed under multipass torsion tests performed at decreasing temperature. In the case of Steels 1, 4 and 5, the specimens were deformed at temperatures in the range of 1180–700°C, with a constant 30°C decrease in the temperature from pass to pass, and interpass times \( t_{ip} = 30 \) s, while Steels 3 and 6 where deformed in the range of 1150–770°C with a constant decrease of 20°C and \( t_{ip} \) ranging from 5 to 30 s. In the case of Steel 2, both types of deformation conditions were applied. The tests were carried out using different passstrains in the range \( e = 0.25–0.4 \) at a strain-rate of 1 s\(^{-1}\). In a particular test, the strain per pass, the strain-rate and the interpass time were held constant. For each one of the tests the non-recrystallization temperature was determined from the Mean Flow Stress (MFS) vs. \( 1/T \) plots using the method proposed by Bai et al. The deformation conditions and the non-recrystallization temperatures determined for each test are shown in Table 3.

The fractional softening which takes place between passes was determined for each one of the tests. As the interpass fractional softening must be determined under anisothermal conditions, the following equation, proposed...
where an interval of $2$. With the selection of these steels, it is possible to study the strain per pass and the interpass time, as well as steel composition (Nb, C and N contents).

### 4. Model Validation

The microstructural model presented was applied to the multipass torsion tests performed with Steels 1 to 6 in Table 2. With the selection of these steels, it is possible to study an interval of $[\text{Nb} \cdot (C + (12/14) \cdot N)]$ product ranging from 0.0015 to 0.004 covering the most usual commercial grades of Nb microalloyed low carbon high strength steels.

In Fig. 5 the evolution of the interpass softening predicted by the model is compared to the experimental results at different deformation conditions: pass-strain $= 0.3$ and interpass-time $= 30 \text{s}$ in the case of Steels 1 and 5, pass-strain $= 0.4$ and interpass-time $= 20 \text{s}$ for Steels 2, 3 and 6 and pass-strain $= 0.4$ and interpass $= 30 \text{s}$ for Steel 4.

The model determines when strain-induced precipitation starts, as well as its effect on subsequent deformation passes. Once precipitation starts after a given deformation pass (the calculated temperature, $T_{\text{nr}}$ is indicated in Fig. 5) the model predicts a more faster fall in the fractional softening in good agreement with the experimental observations. This suggests that the model predictions for the precipitation start temperatures are reasonable, although for Steels 1 and 4 with a higher carbon content, it seems that the calculated precipitation kinetics are somewhat accelerated, resulting in a higher precipitation start temperature. As a consequence, once the onset of precipitation is predicted the calculated softening values are lower than those observed experimentally.

It is interesting to note that for the results shown in Fig. 5 corresponding to Steels 1, 4, 5, and 6 the predicted strain induced precipitation start temperature, $P_{\text{sr}}$, is close to the value of the non-recrystallization temperature, $T_{\text{ur}}$. It is well known that this temperature depends on the competition between solute drag effect and strain induced precipitation. Usually, at low pass-strains and/or short interpass times the recrystallization of the austenite is controlled by the solute drag effect and the $T_{\text{ur}}$ is determined by the amount of recrystallization that can be reached during the interpass interval without interaction with precipitation. However, increasing the strain, the carbon and Nb content, and/or the interpass time, precipitation is able to occur during the interpass interval, producing a strong retardation on recrystallization when the temperature drops below the $T_{\text{ur}}$. Under these conditions the $T_{\text{ur}}$ represents the temperature below which strain induced precipitation starts to take place during the interpass interval. In the aforementioned examples from Fig. 5, relatively large pass-strains and long interpass times are applied, thus the $T_{\text{ur}}$ is caused by strain-induced precipitation and its value is close to $P_{\text{sr}}$.

In contrast, for Steels 2 and 3, which show the lowest values for $\text{Nb} \cdot (C + (12/14) \cdot N)$ product, the predicted precipitation start temperature is below the calculated $T_{\text{ur}}$. The results of the model indicate that for these two steels the $T_{\text{ur}}$ is due to solute drag effect, and that precipitation occurs at lower temperatures.

When shorter interpass times or lower strains are applied in the case of Steel 3, as shown in Figs. 6(a) and 6(b), strain induced precipitation is not predicted to occur during the whole deformation schedule. However, the predicted softening values are close to the experimental ones. These results suggest that under some deformation conditions, solute drag effect may be the most relevant mechanism in retarding static recrystallization without the implication of precipitation. The evolution of softening in this case is characterized by a gradual decrease from pass to pass as the temperature decreases, as can be seen in Figs. 6(a) and 6(b), involving a relatively wide temperature range and high

| Steel | $T_{\text{w0}}$ (°C) | $D_0$ (µm) | $\epsilon$ per pass | $t_p$ (s) | Cooling rate (°C/s) | $T_{p}$ (°C) |
|-------|-----------------|----------|----------------|---------|----------------|------------|
| Steel 1 | 1400 | 1000 | 0.25 | 0.3 | 30 | 1 | 1012 |
|       |       |       |       |       |       | 974 |
| Steel 2 | 1375 | 730 | 0.25 | 0.3 | 30 | 1 | 1015 |
|       |       |       |       |       |       | 985 |
|       | 1350 | 550 | 0.35 | 0.4 | 20 |       | 966 |
| Steel 3 | 1440 | 540 | 0.3 | 20 | 1 | 990 |
|       |       |       |       |       |       | 996 |
|       |       |       |       |       |       | 1004 |
|       |       |       |       |       |       | 985 |
|       |       |       |       |       |       | 969 |
|       |       |       |       |       |       | 970 |
| Steel 4 | 1420 | 805 | 0.3 | 30 | 1 | 1001 |
|       |       |       |       |       |       | 996 |
| Steel 5 | 1400 | 780 | 0.3 | 30 | 1 | 1074 |
|       |       |       |       |       |       | 1046 |
| Steel 6 | 1440 | 585 | 0.4 | 20 | 1 | 1049 |
|       |       |       |       |       |       | 1030 |

by Liu et al., (1981) was used:

$$FS (%) = \frac{\sigma_m^{i-1} - \sigma_m^{i+1}}{\sigma_m^{i-1} - \sigma_m^{i}} \cdot 100 \quad (20)$$

where $\sigma_m^{i}$ and $\sigma_m^{i+1}$ are the maximum and the yield stresses for the $i$-th (at temperature $T_i$) and the $(i+1)$-st (at temperature $T_{i+1}$) passes respectively, while $\sigma_o^{i}$ and $\sigma_o^{i+1}$ are the yield stresses of a fully recrystallized material for the $i$-th and $(i+1)$-st passes. The stresses $\sigma_m^{i}$ and $\sigma_o^{i+1}$ are derived from the pass-to-pass flow curves, while $\sigma_m^{i}$ and $\sigma_o^{i}$ are determined from the relationship derived from the values of the yield stresses measured in the stress–strain curves corresponding to the range of complete recrystallization (initial passes). The yield stresses are determined by the 2% offset method.

It must be pointed out that the evaluation of the interpass softening provides an indirect procedure for the analysis of the strain-induced precipitation kinetics in Nb microalloyed steels, because the evolution of softening during multipass deformation significantly changes depending on the softening retardation mechanism that operates, i.e. solute drag or strain induced precipitation. If solute drag is the principal mechanism, a gradual decrease in the fractional softening from pass to pass as the temperature decreases is usually observed. However, when precipitation takes place interpass softening falls more rapidly with temperature, since precipitation is a stronger retardation mechanism than solute drag. This different behavior in the softening evolution is of course related to deformation conditions, mainly the strain per pass and the interpass time, as well as steel composition (Nb, C and N contents).
number of passes in between complete recrystallization (>85–95% FS) and recrystallization stop (taken as below 20–30% FS). The opposite is seen for example with the softening behavior shown by Steel 5, Fig. 5, where the softening rapidly drops from 95 to 25% (in only 3 passes) due to the early occurrence of strain induced precipitation.

The effect of varying the strain on the predicted and experimental values of softening is shown in Fig. 7. The data corresponds to tests performed with Steel 2, at a fixed interpass time of 30 s and two different values of deformation per pass, 0.25 and 0.35. It can be seen that for the highest applied strain, $\varepsilon=0.35$, the value of the $T_{nr}$ is close to the
predicted precipitation start temperature, while for a strain of $\varepsilon=0.25$ the predicted precipitation temperature is 100°C lower than the $T_{nr}$. Thus, in the first case the value of the $T_{nr}$ is determined by strain induced precipitation, whereas at the lower strain solute drag effect determines the $T_{nr}$.

The results above also indicate that the $T_{nr}$ temperature determined from the MFS vs. $1/T$ plots do not necessarily represent the temperature above which recrystallization is complete (assuming 95% recrystallization). In most cases the amount of softening related to this temperature is lower, showing the lowest values when the deformation conditions and/or the steel composition do not favor the occurrence of strain induced precipitation and recrystallization may only be delayed by solute drag.

Finally, Fig. 8 shows the evolution of the amount of precipitated Nb calculated by the model for different multipass torsion tests. In Fig. 8(a) the evolution of the amount of Nb precipitated for Steels 3 and 6 deformed at the same conditions, $\varepsilon=0.4$ and $t_{ip}=30$ s is shown. These steels have a similar base composition but different Nb additions, 0.029 and 0.06% respectively. As expected, Nb precipitation begins at higher temperatures for the 0.06% Nb steel, and the amount precipitated per pass is also larger. It is interesting to note that at the end of the deformation schedule, for the 0.06% Nb steel, more than half of the Nb is precipitated, whereas, in the low Nb steel, the amount of Nb precipitated is only about 28% of the total Nb.

Figure 8(b) shows the amounts of Nb precipitated in Steel 3 for a constant strain per pass of $\varepsilon=0.4$ and two different interpass times, 20 and 30 s. It can be seen that precipitation starts at slightly higher temperatures for the condition corresponding to 30 s. As the temperature decreases and Nb precipitates, the amount of Nb present in solution is continuously reduced, therefore the driving force for strain induced precipitation decreases and as a result no precipitation is predicted to occur in the last 5 passes at 20 s interpass time and in the last 3 passes at 30 s. As expected, the final amount of precipitated Nb predicted by the model is higher for the longest interpass time of 30 s.

The results in Fig. 8 denote that at the end of deformation a significant amount of Nb can remain in solid solution, this quantity being larger with increased Nb content in the steel and with a reduction in the interpass time. This Nb in solution may precipitate later; during or after phase transformation and can contribute to a strengthening of the steel.4,5) Thus, knowledge of the amount of Nb in solution prior to transformation is a very useful tool for improving the thermomechanical treatments of steel.

5. Conclusions

(1) A model to predict the evolution of the austenite microstructure in Nb microalloyed steels during multipass rolling is proposed. The model takes into account the interaction between static recrystallization, solute drag and strain induced precipitation, the latter playing an important role in this study, allowing the main mechanism involved in retarding recrystallization between passes to be identified in
(2) The kinetics of Nb precipitation were modeled taking the equation proposed by Dutta and Sellars as the base for calculating the precipitation start time. This equation was modified assuming that the constants $A$ and $B$ in the equation had a dependence on steel composition. Such modification improves the applicability of the equation to a wide range of steel compositions and deformation conditions. The calculation of the amount of Nb precipitated after each pass was also implemented in the model.

(3) The results clearly show that under some deformation conditions, for example in the range of low pass-strains and/or short interpass times, solute drag is the most important mechanism in delaying static recrystallization.

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