Mathematical Modeling of the Recrystallization Kinetics of Nb Microalloyed Steels

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The recrystallization behavior of Nb microalloyed steels was studied using hot torsion testing with the aim of modeling the recrystallization processes taking place during hot rolling. Continuous and interrupted torsion tests were performed in the temperature range 850 to 1050°C at strain rates of 0.05 to 5/s on selected low carbon steels containing Cr, Mo, Nb, Ni and Ti. The kinetics of static and metadynamic recrystallization were characterized and appropriate expressions were formulated for the recrystallization kinetics. These are shown to depend on steel composition and the processing conditions. The rate of metadynamic recrystallization increases with strain rate and temperature and is observed to be independent of strain, in contrast to the observations for static recrystallization. By means of extrapolations to mill strain rates, it is shown that metadynamic recrystallization will always be more rapid than static recrystallization, even at the largest possible accumulated strains. These calculations support the view that the unexpected load drops occasionally observed in industrial mills (particularly in the final few passes) are probably due to strain accumulation leading to the initiation of dynamic recrystallization, followed by metadynamic recrystallization.

KEY WORDS: hot rolling; dynamic recrystallization (DRX); static recrystallization (SRX); metadynamic recrystallization (MDRX); critical strain (εc) for DRX; strain accumulation; strain induced precipitation.

1. Introduction

The past two decades have seen a marked increase in the ability to control and optimize the final properties and dimensions of hot rolled products.1–4 These improvements have been brought about through the application of improved control and measurement systems, leading to the explicit control of the microstructural events that occur during rolling and cooling. The approaches taken by various steel companies to achieve this control vary, based upon local hardware, their technical abilities and on the demands of the market. In many cases, this has been at least partly achieved through the use of mathematical models. It is interesting to note, though, that while the use of modeling is a fairly recent phenomenon, most of the theoretical basis for these models has been in existence for over 50 years, the papers by Orowan5 and Sims6 being published in 1943 and 1945, respectively.

While all of these activities have been pursued by most steelmakers to increase productivity and yields, there is one particular area of modeling that is still largely in the developmental stage, with few on-line applications to date. This is the modeling of microstructure evolution during rolling, including the interaction between microstructure evolution and the thermal and mechanical models. The present research deals with the development of certain submodels within this generic area of hot rolling modeling, with particular emphasis on the softening mechanisms and strain accumulation.7–9)

The aim of this study was to produce data that will assist in the computer modeling of steel rolling, as well as in the understanding of the effect of chemistry on the hot rolling process. A systematic study of the effect of different amounts of Cr, Mo, Nb and Ni addition on the activation energies for DRX, SRX and MDRX and on the kinetics of static and metadynamic recrystallization was therefore carried out. Finally, using suitable extrapolations, the rates of softening by SRX and MDRX under strip rolling conditions (ε≈100/sec) are compared.

2. Experimental Procedure

Seven microalloyed steel grades were investigated in this work. One grade was used as a reference, both for the torsion tests and for mill log analysis. The chemical compositions are listed in Table 1.

Two types of torsion tests were performed in this investigation: (i) single pass torsion tests, and (ii) interrupted torsion tests. Torsion test specimens with a gauge section of 22.2 mm length and 6.4 mm diameter were machined from hot rolled transfer bars. The austenitization temperature prior to deformation was 1200°C for all samples. This temperature was selected after calculating the solution temperature using the equation of Irvine et al.10 given by:

\[
\log[Nb] = 12 \frac{C}{14} + 6.770 \frac{T}{T} = 15.226 - 12.057
\]
This equation does not allow for the effects of Mn and Si on Nb(C, N) solubility, but gives a good approximation. According to Eq. (1), the solution temperature for the present Nb grade was 1110°C.

Continuous torsion tests were carried out to determine the peak strain associated with DRX at each experimental temperature in the range 1050–850°C and in the strain rate range $10^{-2}$–$10^{-1}$/s. Interrupted torsion tests were then conducted over the temperature range 1050–950°C, strain rate range $10^{-2}$–$10^{-1}$/s, interpass time range 0.5–500 s, and pass strain range 1/5–3 times the peak strain. This was done so as to evaluate the effects of the deformation variables on static and metadynamic softening. For this purpose, the following technique of interrupted testing was chosen. After reheating the specimen at 1200°C for 15 min, the specimen was deformed at 0.5/s strain rate to 0.5 strain, which is beyond the critical strain for DRX. It was then cooled down to the test temperature, at which double twist deformations were applied under the conditions described below. This entire procedure was repeated 6 times on each sample; i.e. the behavior over the full range of interpass times was determined on a single sample.

Testing was first interrupted at 1/5, 1/4 and 1/3 times the peak strain of DRX, $e_p$, for strain rates ranging from 0.05 to 5/s and temperatures from 1050 to 950°C. After unloading, the specimen was held at the test temperature for times between 0.5 and 500 s. Then, the specimen was reloaded and the same amount of strain was applied. The deformation schedule for the interrupted torsion tests is shown in Fig. 1.

The measured torque $\Gamma$ and twist $\Theta$ were converted to von Mises effective stress ($\sigma$) and strain ($\varepsilon$) using the following equations:

$$\sigma_{eq} = \frac{3.3 \cdot \sqrt[3]{T \cdot T}}{2 \pi^3}, \quad \varepsilon_{eq} = \frac{r \Theta}{L \sqrt{3}} \quad \text{...........(2)}$$

Here, $R$ and $L$ are the gauge radius and length of the specimen, respectively. In order to determine the time for 50% recrystallization, the value of the torque associated with yielding was defined using a 0.2% offset method in the multiple-twist torsion tests.

3. Results and Discussion

3.1. Dynamic Recrystallization

Typical stress-strain curves are shown in Fig. 2 for the first steel (base steel). These were determined at 0.5/s and over the temperature range from 850 to 1050°C. All the curves display a rapid initial rise to a stress maximum, characterized by the peak strain and stress ($e_p$, $\sigma_p$), followed by a gradual fall to a constant stress value (steady state stress, $\sigma_{ss}$). Generally, as the strain rate was increased or the temperature decreased, the peak broadened and the associated stress and strain values increased. Similarly, the steady state stress was raised. This is typical of materials that recrystallize dynamically when deformed at temperatures above half their melting points. The six remaining steels displayed similar behaviors. The initial grain size affects the critical strain for dynamic recrystallization, $e_c$, the peak strain, $e_p$, and the kinetics of DRX. Finer grain sizes lead to lower critical strains and peak strains, as discussed in more detail in Ref. 11. This is because dislocations accumulate more rapidly and the higher grain boundary area per unit volume leads to faster dynamic recrystallization kinetics.

The peak strains for all seven steels are plotted as a function of temperature in Fig. 3 for $\dot{\varepsilon} = 1$/s. Similar results were obtained at 0.5/s and 5/s. The equations specifying these peak strains are listed in Table 2. It can be seen that the peak strain for DRX is significantly altered by the changes in composition. The six steels containing the higher level of Nb (0.06% Nb) display higher peak strains than the low Nb (0.03%) steel (steel 7). Such results show that the effects of composition are attributable to solute drag of

| Chemical compositions of the steels (seven basic microalloyed grades). |
|-----------------|-------|-------|-------|-------|-------|-------|-------|
| C   | Mn   | Si   | Nb   | Mo   | Ti   | Cr   | Remark |
| 1   | 0.068 | 1.47  | 0.202 | 0.056 | 0.158 | 0.014 | Base Composition |
| 2   | 0.067 | 1.50  | 0.205 | 0.058 | 0.283 | 0.015 | Mn effect |
| 3   | 0.070 | 1.50  | 0.205 | 0.059 | 0.161 | 0.015 | Ni effect |
| 4   | 0.070 | 1.50  | 0.206 | 0.059 | 0.156 | 0.015 | Cr effect |
| 5   | 0.074 | 1.51  | 0.206 | 0.061 | 0.157 | 0.015 | Nb effect |
| 6   | 0.072 | 1.49  | 0.207 | 0.062 | 0.155 | 0.015 | Base Composition |
| 7   | 0.071 | 1.50  | 0.205 | 0.032 | 0.151 | 0.015 | Base Composition |

Fig. 1. Deformation schedule for the interrupted torsion tests.

Fig. 2. Stress–strain curves for the reference steel.

Table 1.
the various alloying elements in solution (Cr, Mo, Nb and Ni). There is also a pinning effect of Nb(C, N) precipitates at temperatures below 950°C, although these lower temperatures were not investigated here. In the absence of precipitation, solute drag affects the motion of dislocations and sub-boundaries on the one hand and of grain boundaries on the other. The former events are generally lumped together as recovery (and include nucleation), whereas the latter constitutes recrystallization proper.

The peak strain is strongly influenced by the presence of alloying elements because it is determined by the softening kinetics associated with DRX; this mechanism first balances and then overcomes the work hardening that is taking place. As DRX basically involves the migration of grain boundaries, the addition of Nb, which has a strong solute drag effect, retards the rate of grain boundary motion, shifting the peak strain to the right hand side, i.e. to higher strains. To a lesser extent, other elements, such as Mn and Si, also have their drag effects.

Various empirical equations have been proposed to describe the thermally activated process of hot deformation. Most commonly, the flow stress is related to the strain rate and temperature as follows:

\[ \dot{\varepsilon} = A (\sinh \alpha \sigma)^{n'} \exp \left(-\frac{Q_{\text{def}}}{RT} \right) \] .........(3)

where \( A \), \( \alpha \) and \( n' \) are constants, \( Q_{\text{def}} \) is the activation energy of deformation, \( \dot{\varepsilon} \) is the strain rate, \( \sigma \) is the stress and \( T \) is the absolute temperature. The parameters \( \alpha \), \( n' \) and \( Q_{\text{def}} \) are determined at the peak stress of DRX. To determine the first two quantities, a method similar to that used by Uvira and Jonas\(^ {12}\) was adopted here. It consists of varying \( \alpha \) over a narrow range, which results in a variation in \( n' \). The final value of \( \alpha \) chosen is the one that gives the best correlation coefficient over the full range of temperature and strain rate. The value of \( \alpha \) that gives the best fit is around 0.005/MPa for the seven steels investigated here and the resulting values of \( A \) and \( n' \) are in general agreement with the observations of other workers.\(^ {13,14}\) The full set of \( \dot{\varepsilon}/\sigma \) relations determined here are listed in Table 3.

Using regression analysis, the values of the activation energy for hot deformation were derived; these can be expressed by the following correlation between \( Q \) and the chemical composition:

\[ Q_{\text{def}} = 297 + 641[\% \text{Nb}] + 123[\% \text{Mo}] + 1[\% \text{Ni}] - 111[\% \text{Cr}] \] ..................................................(4)

The activation energy values calculated using this equation are compared with those obtained from the torsion tests in Fig. 4. They are comparable to values found in the literature. It is interesting to note that higher activation energies were observed when the higher alloying element concentrations had stronger retarding effects on DRX.

### 3.2. Static Recrystallization

A selection of interrupted torsion stress–strain curves determined at 950°C, a constant true strain rate of 0.5/s, a pass strain of 25% of the peak strain, and interpass times of 0.5–500 s is displayed in Fig. 5 for the 0.28Mo steel. Several interrupted torsion tests with increasing unloading...
times are plotted together to demonstrate the effect of unloading time on the flow curves. Similar results were obtained on the other materials. From these figures, it can be seen that when the unloading time is short, little softening takes place. As a consequence of the lack of softening, the second twist curve displays less work hardening. Thus, the flow curve resulting from the second twist follows the continuous one fairly closely.

When the unloading time was increased, static recovery occurred and diminished the dislocation density slightly. Under these conditions, the second twist curve displays a "transient" in flow stress, during which the dislocation density is restored to its value just prior to the interruption of straining. If the unloading time is long enough to allow static recrystallization (SRX) to destroy the dislocation structure established during the first deformation, then the second curve work hardens in a similar way to the first curve so as to rebuild the dislocation structure.

The effects of strain rate, temperature and pass strain on softening in the 0.28Mo steel are presented in Fig. 6. Under the present conditions, all the softening curves were of sigmoidal shape with an Avrami time constant $n = 1$ (see below). Increases in the strain rate, temperature and pass strain over the ranges shown accelerate the softening kinetics by almost an order of magnitude. The softening behaviors of the six remaining materials were qualitatively similar. It should be noted that only prestrains of $0.2\varepsilon_p$, $0.25\varepsilon_p$ and $0.33\varepsilon_p$ were used, avoiding prestrains in the range $0.5\varepsilon_p$ to $\varepsilon_p$. This was because static and metadynamic recrystallization can be expected to overlap in the latter range, making it difficult to determine the kinetics of the individual mechanisms. The differences between the kinetics are of particular importance in the present work. This is a topic to which we will return in more detail below.

The effects of the alloying elements can be seen when comparing the softening curves of the seven different steels. It is readily evident that the addition of the alloying elements Cr, Mo, Nb and Ni retards the softening kinetics. In the present work, these retardation effects are due to solute drag as no carbonitride precipitation took place. The small differences in the softening behaviors of the seven steels are consistent with the differences displayed during continuous deformation.

Recrystallization processes involving nucleation and growth can be described by the Avrami equation:

$$X_{SRX} = 1 - \exp \left[ -0.693 \left( \frac{t}{t_{50\%}} \right)^n \right] \quad (5)$$

The recrystallization kinetics can be completely described in this way once $n$ and $t_{50\%}$ are given. In the case of SRX, $n$ has been reported to increase with decreasing temperature and also to decrease from 2 to 1 when the grain size was increased from 104 to 530 µm.\(^{15}\) In the present work, the times for 50% softening were determined by neglecting the stage 1 softening attributable to recovery. The effect of strain rate on $t_{50\%}$ was estimated at constant temperature and strain in this way for the seven steels. $t_{50\%}$ was found to have a power dependence on strain rate:

$$t_{50\%} \propto \dot{\varepsilon}^r \quad (6)$$

where the exponent $r$ has an average value of $-0.43$ for the seven steels. Here we take $n = 1$. The strain rate dependence in the case of metadynamic recrystallization is usually much higher, as will be seen below in more detail. Thus, the static recrystallization kinetics are much less sensitive.
to strain rate than are those of metadynamic recrystallization. This is of especial importance in hot strip and rod mills, where the strain rates are 10 to 100 times higher than the ones used in laboratory tests. The dependencies of static softening on temperature and pass strain were also obtained by the above method (not shown here). In nearly all of these previous studies,\textsuperscript{16,17} the exponent associated with the initial strain size was taken to be $q=2$. Finally, the $t_{50\%}$ expressions were calculated; these are presented in Table 4.

The current work suggests that high levels of Ni have no significant effect on the activation energy for recrystallization in the solute drag regime, although they do retard recrystallization. Previous work by Yamamoto et al.\textsuperscript{18} on decarburised steels showed that a Ni addition of 1 wt% led to a minor increase in the time to a given fraction of recrystallization. However, Maehara et al.\textsuperscript{19} observed a marked increase in the time to 50% softening at 900°C after the addition of Cu and Ni, with the former having effects approaching the levels of retardation achieved by a similar level of Nb. This result is consistent with the large size mismatch between Cu and Fe atoms, which has previously been used to rationalize the retarding influences of various solute elements.\textsuperscript{19} Cu has a similar mismatch to Mo and the retardations observed by Maehara et al.\textsuperscript{20} for the two elements are similar. Ni has a similar atomic size to Fe and would not be expected to retard recrystallization by as much. It should be noted though that Maehara et al.\textsuperscript{20} used a strain rate of only 0.02/s.

Other factors have been suggested as important for predicting the retarding influence of solutes on recrystallization. These include the electronic structure of the atom\textsuperscript{20} and the relative velocities of the migrating interface and the solute.\textsuperscript{21} Given the available information on these parameters, it would be expected that Cr, Mn and Ni would have less effect on recrystallization (per atomic percent of addition) than Mo or Nb, for example.

It should be remembered that the current steels already contained Nb and that the effect of multiple additions has not been extensively studied, except in the work by Akben et al.\textsuperscript{20} In their research, it appeared that certain combinations of elements, particularly Mo with Nb, led to synergistic increases in the retardation of recrystallization. In the current study, only combined additions have been studied, with the constant presence of Nb. As already mentioned above, the results obtained in this way will be compared and considered in more detail below.

### Table 4. $t_{50\%}$ times for static recrystallization.

<table>
<thead>
<tr>
<th>Steel</th>
<th>Equation</th>
</tr>
</thead>
<tbody>
<tr>
<td>Steel 1 (base)</td>
<td>$t_{0.5%\text{SrX}} = 2.1 \times 10^{16} \cdot 4.4 \cdot e^{3.1} d_f^2 \exp\left(\frac{264000}{R T}\right)$</td>
</tr>
<tr>
<td>Steel 2 (0.28 Mo)</td>
<td>$t_{0.5%\text{SrX}} = 6.0 \times 10^{17} \cdot 4.4 \cdot e^{3.1} d_f^2 \exp\left(\frac{283000}{R T}\right)$</td>
</tr>
<tr>
<td>Steel 3 (0.21 Ni)</td>
<td>$t_{0.5%\text{SrX}} = 3.3 \times 10^{16} \cdot 4.4 \cdot e^{3.1} d_f^2 \exp\left(\frac{261000}{R T}\right)$</td>
</tr>
<tr>
<td>Steel 4 (0.50 Ni)</td>
<td>$t_{0.5%\text{SrX}} = 7.8 \times 10^{16} \cdot 4.4 \cdot e^{3.1} d_f^2 \exp\left(\frac{253000}{R T}\right)$</td>
</tr>
<tr>
<td>Steel 5 (0.30 Cr)</td>
<td>$t_{0.5%\text{SrX}} = 7.6 \times 10^{15} \cdot 4.4 \cdot e^{3.1} d_f^2 \exp\left(\frac{230000}{R T}\right)$</td>
</tr>
<tr>
<td>Steel 6 (0.61 Cr)</td>
<td>$t_{0.5%\text{SrX}} = 5.8 \times 10^{14} \cdot 4.4 \cdot e^{3.1} d_f^2 \exp\left(\frac{207000}{R T}\right)$</td>
</tr>
<tr>
<td>Steel 7 (0.03Nb)</td>
<td>$t_{0.5%\text{SrX}} = 4.0 \times 10^{16} \cdot 4.4 \cdot e^{3.1} d_f^2 \exp\left(\frac{246000}{R T}\right)$</td>
</tr>
</tbody>
</table>

*The grain size dependence is taken from Ref. 15.

### 3.3. Metadynamic Recrystallization

The interrupted stress–strain curves obtained at 1 000°C and a strain rate of 0.5/s with varying interpass times, are presented in Fig. 7 for the 0.28Mo steel. The double twist curves associated with the other grades were again qualitatively similar. Here, the pass strain was equal to the peak strain so that the effects of the deformation variables on metadynamic softening could be evaluated. It can be seen that much softening takes place, even when the unloading time is quite short. As a consequence of the rapid softening, the second curve displays renewed work hardening. Metadynamic recrystallization occurs by the continued growth of the nuclei formed as a result of dynamic recrystallization during prestraining. Hence the operation of MDRX does not require an incubation time and such rapid interpass softening can affect the mechanical properties, even when the interpass times are short. The second curve work hardens in a similar way to the first curve so as to rebuild the dislocation structure; the large amount of softening taking place during the interruption is responsible for the presence of a new dynamic recrystallization peak in the second curve. This peak appears at larger strains when the unloading time is increased due to the necessity of rebuilding more of the dislocation structure before dynamic recrystallization can be resumed.

The effects of strain rate and temperature on metadynamic softening are presented in Fig. 8 for the 0.28Mo steel. All the softening curves are of sigmoidal shape; a rapid stage, which reaches 40 to 50%, can be identified. At a strain rate of 5/s or temperature of 1 050°C, the softening process is so fast that the beginnings of the process cannot be detected, even after the shortest achievable interruption time of 0.5/s. Nevertheless, softening did not go to completion at low temperatures, probably due to precipitation. At higher strain rates, the driving force for metadynamic recrystallization is expected to be still higher. Under these conditions, the half-recrystallization time can be estimated to be reduced to below 100 ms. The results obtained on the other steels are not presented here to conserve space but are again similar.

The factors affecting metadynamic softening differ markedly from those influencing the static recrystallization results described in the previous section for the same steels. Here the rate of recrystallization is relatively unaffected by
temperature and composition, while strain rate has a significant effect. This may arise because the dislocation density under conditions of DRX is much more sensitive to strain rate (via the subgrain size) than when straining is interrupted before DRX has been initiated (i.e. when \( e < e_i \)).

The effect of strain rate on \( t_{50\%} \) was estimated at constant temperature and strain for the seven steels. \( t_{50\%} \) was found to have a power dependence on strain rate (Eq. 6), where the exponent \( r \) is about \( 0.84 \) for the seven materials. This value is about twice that for SRX (see Table 4). The effect of temperature on \( t_{50\%} \) was also estimated at constant strain rate and strain for the seven grades. The logarithm of \( t_{50\%} \) was found to be proportional to the inverse temperature. This dependence is characteristic of a thermally activated process, which can be represented as follows:

\[
Q_{\text{RT}} t_{50\%} \propto \exp \left( \frac{Q_{\text{app}}}{RT} \right)
\]

where \( Q_{\text{app}} \) is the apparent activation energy of the process. The temperature dependence of metadynamic recrystallization was found to be distinctly lower than that of static recrystallization. Finally, the \( t_{50\%} \) expressions were calculated and are presented in Table 5.

The values of \( Q_{\text{MDRX}} \) determined in the current work are approximately 170 kJ/mol. This overall activation energy indicates that decreasing the temperature will lead to smaller increases in the time to a certain level of softening than in case of static recrystallization. One possible interpretation of the approximate 3:2 ratio of \( Q_{\text{SRX}}/Q_{\text{MDRX}} \) is that the former is associated with the activation energies of both nucleation and growth whereas the latter, given that nucleation has already taken place during straining, only involves the activation energy for growth. In addition to the \( Q \) values displayed in Table 5, the pre-exponential A values are also important. This is because both are required to describe the metadynamic recrystallization kinetics.

It is now of interest to compare the kinetics of static and metadynamic recrystallization. This comparison was undertaken for all seven steels. The differences between the two types of kinetics are presented in the form of plots of the time for 50% recrystallization in Fig. 9 for the 0.28% Mo steel. The two sets of curves are broken at the critical strain for dynamic recrystallization. At strains below the critical strain, softening is due solely to static recrystallization. The time for 50% recrystallization is highly sensitive to the pre-strain. When deformation is interrupted beyond the critical strain, metadynamic recrystallization begins to take place. Its kinetics are no longer dependent on strain but depend strongly on the strain rate. This abrupt change in the recrystallization kinetics is of considerable importance in the modeling of hot rolling. Due to the high strain rates involved, e.g. 50 to 200/s in hot strip mills, MDRX is expected to take place rapidly, i.e. in less than 0.5 s, as estimated by extrapolating the present data to higher strain rates.

### 3.4. Effect of Alloying Elements on the Activation Energy and on \( t_{0.5\%} \)

The effect of the alloying elements on the activation energies for DRX, SRX and MDRX is illustrated in Fig. 10. It should be noted that the addition of Nb and Mo leads to increases in the activation energies while that of Cr has a decreasing effect.
The times for 50% static and metadynamic recrystallization, obtained by smoothing the softening curves presented in Figs. 6 and 9 (and employing the equivalent curves obtained on the six other steels), are illustrated in Figs. 11(a) and 11(b). From these figures, it can be seen that $t_{0.5}$ increases with the solute content of the steel. The retardation of recrystallization over the temperature range of these figures is due to a solute drag effect, as solute elements have been shown to retard the migration of grain boundaries by segregating to their boundaries. However, the possible precipitation of Nb during specific interpass intervals at temperatures below 930°C could involve retardations due to a combination of solute drag and strain-induced precipitation.

An important trend that can be discerned from Fig. 11(a) is the progressive retarding effect of the alloying elements in solution, mentioned above. As the solute content of the steels is increased from its lowest (0.03% Nb) to its highest (0.61% Cr) level, the rate of the recrystallization is slowed down by a factor of about 4. This is a significant difference, which can play an important role in some types of steel rolling.

Similar remarks apply to Fig. 11(b), except that in this case there is no effect of pass strain, as these results apply to postdynamic recrystallization. The retarding effect of the solute elements is seen to be larger under these conditions than for SRX, amounting to a factor of about 6 in passing from the 0.03% Nb to the 0.50% Ni and 0.61% Cr steels. It should also be noted that the $t_{0.5}$'s illustrated in Fig. 11(b) were determined at a strain rate of 0.5/s. At hot strip mill strain rates of about 100/s, the applicable times are expected to be shorter by a factor of about 100.

The present results are plotted in a more conventional manner in Fig. 12. Here, the range of hot strip rolling temperatures considered was 1000–870°C. Although the time for 50% recrystallization at 870°C is estimated to be about 250–600 s for SRX, it is only 1–7 s for MDRX (at a laboratory strain rate of 0.5/s). After correction for mill strain rates of about 100/s using the equations of Tables 4 and 5 (Fig. 13), these two ranges become 8–50 s and 0.02–0.1 s, respectively. As the interpass times in the finishing stands are as short as 0.5–2 s, it is evident that there is insufficient time for SRX to take place. By contrast, the initiation of
DRX, followed by MDRX, can produce enough softening to cause load drops.

There are three important comments that must be made here regarding Figs. 12 and 13. The first involves the effect of temperature. As \( T_{\text{MDRX}} \), \( T_{\text{DRX}} \), and \( T_{\text{SRX}} \) decrease, the lower the temperature in the finishing range, the greater the expected difference in \( t_{0.5} \) times for MDRX and SRX, as can be seen in the diagrams. The second point concerns the effect of strain rate. Data such as those reported here are measured at laboratory strain rates, which are of course much lower than mill strain rates. Thus, when the differences between \( t_{0.5} \) (MDRX) and \( t_{0.5} \) (SRX) as determined in the laboratory are small, they will always be greater when extrapolated to mill strain rates. (This is because \( r = -0.43 \) for SRX whereas it is \(-0.84 \) for MDRX.)

The final point concerns the effect of strain and arises because the boundary between the accumulated strains leading to MDRX on the one hand or to SRX on the other are only very imprecisely known. This is why the extrapolated \( t_{0.5} \)'s for SRX in Fig. 13 are shown for prestrains up to the peak strain, which is the largest conceivable prestrain that could possibly produce SRX instead of MDRX. It is clear from Fig. 13 that, whether the cut-off for SRX occurs at 0.5\( \varepsilon_p \) (as has been suggested for Nb steels\(^1\leq 3\)) or 0.75\( \varepsilon_p \) (as has been suggested for plain C steels\(^4\)), or even \( \varepsilon_p \), there is a clear demarcation between the \( t_{0.5} \)'s applicable to MDRX and SRX, the former always being distinctly shorter. The present data thus provide important evidence for the interpretation that the unexpected load drops that occur under industrial hot strip mill conditions are due to strain accumulation leading to the initiation of DRX followed by MDRX.\(^{24}\)

4. Summary and Conclusions

In the present work, static, dynamic and metadynamic recrystallization were investigated in seven microalloyed steels. Continuous deformation and interrupted deformation were employed at different temperatures and strain rates.

The peak strains were defined from the torsion flow curves and the values determined in this way increased with Zener-Hollomon parameter in the normal manner. The kinetics of static recrystallization were measured and the rates were found to increase rapidly with strain and temperature. In the case of metadynamic recrystallization, the strain rate dependence was determined and found to be about twice as high as for static recrystallization. The activation energies were measured and observed to be distinctly lower. These dependences differ sharply from those of static recrystallization and contribute to the definition of metadynamic recrystallization as being a distinct post-deformation process.

The kinetics of static and metadynamic recrystallization determined on the present seven steels have been compared. The results indicate that SRX is about two orders of magnitude slower than MDRX at finish rolling temperatures. The data obtained reveal the relatively strong retarding effect of elements in solution on both types of recrystallization. Increasing additions retard recrystallization by factors of up to 4 or 5 in going from the least alloyed (0.03% Nb or 0.06% Nb plus 0.5% Mn) to the most alloyed (0.06% Nb plus 0.50% Ni or 0.61% Cr) grade.

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