Analysis of work hardening and recrystallization during the hot working of steel using a statistically based internal variable model

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Abstract

A mathematical model has been developed which describes the hot deformation and recrystallization behavior of austenite using a single internal variable: dislocation density. The dislocation density is incorporated into equations describing the rate of recovery and recrystallization. In each case no distinction is made between static and dynamic events, and the model is able to simulate multideformation processes. The model is statistically based and tracks individual populations of the dislocation density during the work-hardening and softening phases. After tuning using available data the model gave an accurate prediction of the stress–strain behavior and the static recrystallization kinetics for C–Mn steels. The model correctly predicted the sensitivity of the post deformation recrystallization behavior to process variables such as strain, strain rate and temperature, even though data for this were not explicitly incorporated in the tuning data set. In particular, the post dynamic recrystallization (generally termed metadynamic recrystallization) was shown to be largely independent of strain and temperature, but a strong function of strain rate, as observed in published experimental work.

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Keywords: Metadynamic recrystallization; Stress–strain relationship; Zener–Hollomon parameter

1. Introduction

The thermomechanical processing of steel at elevated temperatures involves a number of microstructural changes. During the actual deformation and between intervals of deformation there is a complex process of work hardening, recovery and recrystallization. Both the recovery and recrystallization restoration processes can occur dynamically (i.e. during deformation) or statically (i.e. between intervals of deformation). All of these processes are strongly influenced by the behavior of the dislocation population and the effect of this on the nucleation and mobility of new strain free grains.

The industrial processing of steels via controlled thermomechanical processing requires accurate knowledge, and from this control, of each of these processes. While earlier efforts concentrated on detailed experimentation of single microstructural events, there are now several groups attempting to develop complex mathematical models that describe both the processing parameters (e.g. thermal and mechanical models) and the evolving microstructure and, through this, the interrelationships between the two. The complexity of the microstructure evolution models varies significantly.

One approach has been to construct models based on the dislocation dynamics during and after deformation, although the majority of work to date of this nature has been limited to describing the work hardening rate during deformation \cite{1-4}. These models have been evaluated for their applicability to hot rolling of steels in the austenitic condition with varying levels of agreement. In all cases, though, these models fail beyond a critical strain due to the occurrence of dynamic recrystallization, which has not been accounted for.

In some cases, simple models have also been developed for dynamic recrystallization, for example, by coupling the dislocation work hardening model to an
empirical equation such as an Avrami law to describe the kinetics of dynamic recrystallization [5–7]. Such an approach provides a reasonable description of the stress–strain curve during dynamic recrystallization beyond the peak strain of dynamic recrystallization, although the physical basis for this description is weak.

When deformation is interrupted, softening takes place at a rate, which depends on whether deformation was interrupted before, or after, the initiation of dynamic recrystallization. In the first case, softening is referred to as static recrystallization and in the latter as metadynamic recrystallization. The mechanisms and resulting kinetics of these processes have been studied and modeled for some time. From a practical point of view it is possible to model this using simple Avrami type laws, with appropriate time constants.

In the following a statistical approach has been taken in modeling the evolving microstructure during and after deformation. As will be shown the model contains a large number of simplifications of the envisaged physical processes, in many cases this is due to the current lack of detailed knowledge in the literature. The model has been applied to a wide range of deformation conditions for austenite and then compared with available data. This includes work hardening and dynamic, static and metadynamic recrystallization.

2. Description of the model

The model is based on the general approach proposed by Sandstrom and Lagneborg [8,9] who developed a series of equations describing the evolution of the dislocation density as a function of the deformation conditions. As described in the following section, aspects of this have been modified in the light of more recent developments. There are also a number of limitations and assumptions, which are detailed in a later section. One of the main differences in the current work, compared with previous approaches, is that the evolution of various dislocation density populations is discretely followed throughout the entire deformation process. In this way a range of internal states can exist throughout the structure and the subsequent behavior of each of these populations can be handled. The model and calculation structure has been described in detail elsewhere [10,11] and the following only summarizes the important elements and improvements.

2.1. Summary of the model

The model describes the evolution of the dislocation populations during and after deformation, based on the approach suggested by Sandstrom and Lagneborg [8,9]. The overall equation describing the evolution of the dislocation population is:

\[
\frac{dG(\rho, t)}{dt} = \phi(\Delta\varepsilon) - g(\varepsilon) \frac{\gamma}{D} m \tau \rho(\rho, t)
\]

where, \( \phi(\Delta\varepsilon) \) represents the athermal storage (hardening) of dislocations, \( g(\varepsilon) \) the thermally activated softening (recovery), while the final term represents, the changes caused by recrystallization, with \( \tau = \mu b^2/2 \). The basic equations are given in Table 1, while Table 2 lists, models used to describe certain variables in these equations. During deformation, strain hardening is controlled by the competition between the storage (equation (2) in Table 1) and annihilation of dislocations (equation (3)) [2,4]. Beyond a critical dislocation density, the recrystallized volume fraction described by (4) increases and contributes to the removal of dislocations. The fluctuation in the dislocation density due to storage, recovery and recrystallization are described by the evolution of the volume distributions of dislocations \( G(\rho, t) \) which has a dislocation density between \( \rho \) and \( \rho + d\rho \) at the time \( t \), where \( G(\rho, t) \) is the distribution function. Eq 1 is solved for each interval of the discretized dislocation density, together with the equations describing grain refinement (equation (5)) and grain growth (equation (6)).

Adiabatic heating during deformation is also taken into account.

The recovery term, used by Sandstrom and Lagneborg, assumes that the recovery events are taking place by the climb of dislocations [8]. However, using this assumption with an activation energy for recovery of 400 kJ mol\(^{-1}\) and a mobility coefficient of \( 5.7 \times 10^7 \) m\(^2\) (Ns\(^{-1}\))\(^{-1}\), led to a significant underprediction of the extent of recovery at low temperatures. A similar expression to (eq.3) has been used by Bergstrom and Aronson where \( k_2 \) was considered as a measure of the probability for remobilization or annihilation of immobile dislocations [1]. In the model by Estrin and Mecking for recovery, this was assumed to be controlled by self diffusion [4]. Previous work on microalloyed steels has shown that the models of Estrin and Mecking [4] or Bergstrom and Aronson [1] predicted the work hardening rate over a wide range of hot rolling conditions [3].

Once deformation is interrupted the storage of dislocations ceases and the evolution of the dislocation density is controlled by the recovery and recrystallization terms in equation (1). The equations for dislocation population, grain refinement and grain growth are then
Table 1
Basic model equations

<table>
<thead>
<tr>
<th>Process</th>
<th>Variables</th>
<th>Direction</th>
<th>Condition</th>
<th>References</th>
<th>Equation</th>
</tr>
</thead>
<tbody>
<tr>
<td>Hardening</td>
<td>$\Delta \rho$, $\rho$</td>
<td>$\rightarrow$</td>
<td>$t &gt; 0$</td>
<td>[8,9]</td>
<td>$d\rho/dt = 1/\beta l$ (1)</td>
</tr>
<tr>
<td>Recovery</td>
<td>$\Delta \rho$, $\rho$</td>
<td>$\rightarrow$</td>
<td>Always</td>
<td>[2,4]</td>
<td>$d\rho/dt = \kappa_0 - \kappa_1 \rho$ (3)</td>
</tr>
<tr>
<td>Recrystallization</td>
<td>$G$, $X$, $\rho$</td>
<td>$\rightarrow$</td>
<td>$\rho &gt; \rho_{cr}$</td>
<td>[8,9]</td>
<td>$dX/dt = v_i \bar{D_v} \rho_0$ (4)</td>
</tr>
<tr>
<td>Grain refinement</td>
<td>$D$</td>
<td>$\rightarrow$</td>
<td>$\rho &gt; \rho_{cr}$</td>
<td>[8,9]</td>
<td>$dD/dt = -DdX/dt \ln N$ (5)</td>
</tr>
<tr>
<td>Grain growth</td>
<td>$D$</td>
<td>$\rightarrow$</td>
<td>Always</td>
<td>[8,9]</td>
<td>$dD/dt = \sigma m_{\mu}/D \exp(-Q_n/RT)$ (6)</td>
</tr>
</tbody>
</table>

Table 2
Equations used to predict the critical variables

<table>
<thead>
<tr>
<th>Variable</th>
<th>symbol</th>
<th>Ref.</th>
<th>Equation</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cell size (free path)</td>
<td>$l$, $d_i$</td>
<td>[12]</td>
<td>$l = d_i = K_1/Z^n$ (7)</td>
</tr>
<tr>
<td>Mobile fraction of boundary</td>
<td>$\gamma$</td>
<td></td>
<td>$g = (0.1 + X)^n(1 - X)(\rho/\rho_{cr})$ (8)</td>
</tr>
<tr>
<td>Number of new grains per old</td>
<td>$N$</td>
<td>[10,11]</td>
<td>$N = 4\pi(D/d_i)^3$ (9)</td>
</tr>
<tr>
<td>Critical dislocation density</td>
<td>$\rho_{cr}$</td>
<td>[12]</td>
<td>$\rho_{cr} = 8\sigma_g/\tau l$ (10)</td>
</tr>
</tbody>
</table>

solved as a function of time for each interval of dislocation density.

Also shown in Table 1 is the order of calculation, which must be performed (termed the direction in Table 1). For example, in hardening (equation (2)) it is necessary to first calculate the hardening of the dislocation population with the highest dislocation density, otherwise there will be positive feedback in the calculation. For recrystallization the order of dislocation densities in which the calculation is performed is unimportant as this process moves the dislocation density from its current value to zero and so will not impact on neighboring dislocation density intervals.

2.2. Model limitations

This model has a number of limitations. Some of these include:

1) The use of a single dislocation type to describe the internal state of the material during deformation and recrystallization. This is an obvious simplification of a much more complex process. For example, others [8] have suggested that there are two major dislocation populations with different rates of evolution and effects on the microstructure. It has been proposed in that work that the dislocation density inside the subgrains governs the flow stress, whereas the dislocation density within the subgrain wall governs the amount of stored energy for the occurrence of recrystallization. However, it has been recently discussed [14] that this may not be the case and that the internal dislocation density also plays a significant role in the rate of recrystallization. If this holds then the error associated with the use of a single dislocation density in both the stress and microstructure evolution equations will be reduced.

2) No information regarding misorientations across grain and subgrain boundaries. The amount of stored energy, and the driving force for recrystallization, is also dependent on the angle of misorientation across boundaries. The heterogeneity in the distribution of stored energy and the spread in misorientation increases with the development of sub-boundaries and is expected to influence the nucleation and growth of recrystallization.

3) The assumption that the subgrain structure is established at an early stage of deformation. The subgrain size is initially dependent on $\sqrt{\rho}$ and then reaches a steady-state size given by an empirical equation related to the Zener–Hollomon parameter [15,16]. Improvement of the model would require a physical model, which describes the subgrain evolution during deformation and interpass times.

4) The model does not take into account the different orientations within the material and its consequence on the mechanical (i.e. stress–strain) behavior. It assumes that all grains follow the same hardening law and that one grain size is associated with one dislocation density.

5) Finally, the model assumes that all evolution equations of dislocation population and grain size are the same during and after deformation. Hence, the mechanisms that take place during dynamic and static recovery or during dynamic, static or metastable recrystallization are assumed to be the same.

3. Determination of material constants

With the current state of knowledge it is difficult to determine the appropriate material constants in the dislocation models described above. Most work related to the development of these more physically based models, has been generated using controlled alloy systems, far removed from commercial steels. In the current work, the approach has been to fit the para-
meters using available experimental data for the stress–strain behavior and the recrystallization kinetics from two independent research groups.

The material constants were obtained using an optimization technique described elsewhere [11,17]. The objective function was defined as a sum of the square of differences between the measured and calculated values of the considered parameters. The optimization technique was performed for a plain carbon steel (in the hot worked in the austenite region) using stress–strain curves from [5] and softening results from [18,19]. The grain boundary energy $\sigma_g$, the shear modulus $\mu$ [20] and $v$ [8] were taken from the literature. The material constants $K_d$, $q$, $k_{20}$, $m_0$, $Q_m$, $q_1$, $q_2$, $p_{PCR}$, $\alpha$ were calculated by optimizing the objective function, starting from values obtained from the literature (Table 3).

The optimization yielded the values shown in Table 4. The error between the predicted and experimental stresses and the time of recrystallization after deformation was 13%. The stress–strain curves at a strain rate of 0.2 s$^{-1}$ at 1100 and 1000 °C were accurately predicted (Fig. 1a). Below 1000 °C, the predicted strain at the peak stress was larger than observed. At a strain rate of 2 s$^{-1}$, the predicted stress was lower at high temperatures, but higher at lower temperatures (Fig. 1b). It appears that little, or no, dynamic recrystallization was predicted by the model below 900 °C.

The strain, temperature and strain rate sensitivity of the predicted softening curves compared reasonably well with the data used to tune the model (Fig. 2) except for a strain of 0.8 where the experimental data were faster. The model predicted a slightly higher temperature sensitivity as the experimental kinetics were slower than those predicted (Fig. 2b).

However, considering the amount of data and the fact that these came from different research groups with different steels and test methods, the overall level of agreement is highly encouraging.

### 4. Analysis of the model

In the following sections various aspects and predictions of the model are examined. Here it is possible to manipulate the various structure equations to examine their effect on the evolution of the total microstructure and the resulting stresses.

#### 4.1. Behavior during deformation

The structure of the model allows the relative influences of dynamic recovery and recrystallization on the stress–strain curve to be investigated (Fig. 3). When only recovery is allowed to take place during deformation, the work hardening rate increases to a saturation strain larger than the experimentally observed peak stress, as expected. When only recrystallization contributes to softening, the stress increases to higher levels than the experimental values before the initiation of dynamic recrystallization. Recrystallization then produces a multiple peak behavior with the first peak located at a lower strain than observed experimentally. Recovery, therefore, lowers the work hardening rate and also retards recrystallization. This raised some interesting considerations for dynamic recrystallization. Multiple peak stress–strain curves are only experimentally observed at low values of the Zener–Hollomon parameter (i.e. low strain rates and/or high temperatures)$^1$. These conditions are also those which favor high rates of dynamic recovery. However, the model actually suggests that dynamic recovery retards this oscillating stress–strain behavior. The significance of this has not been pursued in the current work nor has the performance of the total model been critically examined at low $Z$. These aspects require further investigation.

The retardation of dynamic recrystallization can be further shown by following the effect of recovery on the initiation of dynamic recrystallization. The rate of recovery is controlled by the activation energy $Q_s$ and the constant $k_{20}$. Changing the rate of recovery influences the strain at which 5% dynamic recrystallization is observed. For lower $Q_s$ and higher $k_{20}$, i.e. for a higher recovery rate, dynamic recrystallization becomes more difficult to initiate (Fig. 4) as expected.

One important observation from the current model was that dynamic recrystallization proceeded at the same rate, irrespective of whether there was a critical dislocation density for nucleation, or not (i.e. nucleation assumed to be satisfied at zero strain). Essentially, the model predicted a growth dominated process. As no critical dislocation density was needed to activate

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$^1$ The conditions, which have been suggested to produce this change have been discussed in detail elsewhere [21] and are related to the nucleation density.

### Table 3

Material constants used as starting values

<table>
<thead>
<tr>
<th>Variable</th>
<th>$b$ (m)</th>
<th>$k_2$</th>
<th>$Q_s$ (kJ mol$^{-1}$)</th>
<th>$m_0$ (m$^3$ (Ns)$^{-1}$)</th>
<th>$Q_m$ (kJ mol$^{-1}$)</th>
<th>$v$</th>
<th>$q$</th>
<th>$\alpha$</th>
<th>$\sigma_g$ (J m$^{-1}$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>References</td>
<td>[20]</td>
<td>[13]</td>
<td>[13]</td>
<td>[12]</td>
<td>[12]</td>
<td>[20]</td>
<td>[12]</td>
<td>[2]</td>
<td>[12]</td>
</tr>
<tr>
<td>Value</td>
<td>0.25 × 10$^{-9}$</td>
<td>30</td>
<td>25</td>
<td>3.5 × 10$^4$</td>
<td>360</td>
<td>6</td>
<td>0.167</td>
<td>1</td>
<td>0.8</td>
</tr>
</tbody>
</table>
recrystallization, the evolution of a critical strain, defined as the strain for 5% dynamic recrystallization, was examined. The critical strain was found to increase with strain rate and decrease with increasing temperature (Fig. 5). This behavior is similar to that observed experimentally for dynamic recrystallization. The ratio of the critical strain over the peak strain of about 0.6 was comparable to the 0.6–0.85 reported in the literature based on optical microscopy, or a change in the work hardening rate [21,22]. The lower ratio may be due to the difficulty of experimentally establishing the critical strain of dynamic recrystallization. This ratio also depends on the alloying content and the processing conditions, although it was interesting to note that the model predicted little variation in this ratio over a wide range of temperatures and strain rates.

4.2. Post deformation behavior

Once the deformation is interrupted, softening takes place during the interpass time. There is only a small amount of recovery and the major softening event is recrystallization. The recrystallization taking place after unloading at a strain less than the critical strain to initiate dynamic recrystallization is referred to as static

<table>
<thead>
<tr>
<th>Variable</th>
<th>$K_d$ (m s$^{-1}$)</th>
<th>$q$</th>
<th>$k_{20}$</th>
<th>$Q_s$ (kJ mol$^{-1}$)</th>
<th>$m_o$ (m$^3$ (Ns)$^{-1}$)</th>
<th>$Q_m$ (kJ mol$^{-1}$)</th>
<th>$q_1$</th>
<th>$q_2$</th>
<th>$x$</th>
</tr>
</thead>
<tbody>
<tr>
<td>Value</td>
<td>$0.26 \times 10^{-3}$</td>
<td>0.1</td>
<td>34</td>
<td>20</td>
<td>$1.4 \times 10^5$</td>
<td>340</td>
<td>0.5</td>
<td>2</td>
<td>1.23</td>
</tr>
</tbody>
</table>

Table 4
Material constants determined by optimization

Fig. 1. Stress–strain curves predicted at (a) 0.2 s$^{-1}$ and (b) 2 s$^{-1}$ for the temperatures 800–1100 °C. The experimental curves are described by selected points.

Fig. 2. Comparison of the experimental and predicted softening curves, (a) effect of strain and (b) effect of temperature.
recrystallization, as opposed to metadynamic recrystal-
lization when interrupting beyond the critical strain for
dynamic recrystallization. The distinction has been
based on the different types of recrystallization
kinetics observed for static and metadynamic recrystallization
[23,24]. The dependence of the recrystallization kinetics
in the model on strain, strain rate and temperature was
evaluated by comparing the change in time for 50%
recrystallization as a function of the deformation
conditions (Fig. 6).

As the interruption strain is increased, the time for
50% recrystallization decreases and then saturates. The
change in the time for 50% recrystallization with strain
is smooth and continuous which differs from some
suggestions of a discontinuity at the critical strain
[22,23]. At low strain, the time for 50% recrystallization
decreased slightly with an increase in the strain rate (Fig.
6a). Beyond the strain at which dynamic recrystallization
is initiated the time for 50% recrystallization decreased by almost an order of magnitude with an
order of magnitude increase in strain rate. This change

Fig. 3. Effect of dynamic recovery (square), dynamic recrystallization (round), dynamic recovery and recrystallization (triangles) on stress
strain curves compared with the experimental data (stars) for a strain rate of 0.2 s\(^{-1}\) and temperatures of (a) 900 °C and (b) 1000 °C.

Fig. 4. Effect of the recovery parameters \(Q_s\) and \(k_{20}\) on the initiation of
dynamic recrystallization.

Fig. 5. Evolution of the strain for 5% dynamic recrystallization and
the peak strain with (a) temperature and (b) strain rate.

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Fig. 5. Evolution of the strain for 5% dynamic recrystallization and
the peak strain with (a) temperature and (b) strain rate.
in the strain rate sensitivity of recrystallization with strain has been observed experimentally [22–25].

Temperature was also found to affect the time for 50% recrystallization. For small strains an increase in temperature increases the rate of recrystallization. However, at larger strains (ε > εc) the time for 50% recrystallization becomes insensitive to temperature. These trends are in agreement with the observations of Hodgson et al. [18,24] who reported no effect of temperature on the kinetics of metadynamic recrystallization after unloading in the steady state of dynamic recrystallization. Roucoules et al. [25] also observed a much weaker temperature dependence for metadynamic recrystallization than for static recrystallization.

The time for 50% recrystallization can usually be empirically related to the strain rate, strain and temperature as:

\[ t_{50\%} = A \dot{\varepsilon}^n \varepsilon^m \exp \left( \frac{Q}{RT} \right) \]  

(12)

Table 5
Coefficients of the empirical \( t_{50\%} \) equations

<table>
<thead>
<tr>
<th></th>
<th>A</th>
<th>n</th>
<th>m</th>
<th>Q (kJ mol(^{-1}))</th>
</tr>
</thead>
<tbody>
<tr>
<td>SRX</td>
<td>1.11 × 10(^{-11})</td>
<td>−2.65</td>
<td>−0.33</td>
<td>218</td>
</tr>
<tr>
<td>MDRX</td>
<td>0.14</td>
<td>0.81</td>
<td>−0.87</td>
<td>23</td>
</tr>
</tbody>
</table>

The present model was used to generate coefficients for (12) (Table 5). The strain rate exponent of −0.33 for static recrystallization is between the −0.28 of Choquet et al. [26] and the −0.41 of Laasraoui et al. [27]. The apparent activation energy is close to the 250 kJ reported by Hodgson [18,24] and Laasraoui et al. [27], but less than the 300 kJ mol\(^{-1}\) of Sellars [22]. The strain exponent is also in agreement with the −2.5 of Hodgson [18,24] but lower than the −3.8 of Laasraoui et al. [27]. For metadynamic recrystallization the strain rate exponent was close to the −0.8 reported by Hodgson et al. [24] and −0.61 Roucoules et al. [25], while the activation energy was less than that reported by Roucoules et al. [25]. The strain exponent was slightly positive in the metadynamic regime, whereas most previous work suggests that there is no sensitivity to strain [24].

5. Discussion

The above has demonstrated that the model can at least qualitatively, and in many cases quantitatively, predict the evolution of stress and recrystallization using this simple, single internal variable model. Hence, even though some of the approximations appear to represent very significant simplifications of the real physical processes, the model appears to provide a powerful tool for trying to infer some of the dominant microstructural processes that are operative during the hot working of steel. One of the most significant aspects of this work has been the ability of the model to predict the change in static recrystallization behavior when moving from an initially work hardened microstructure, to one which contains a mixture of work hardened and dynamically recrystallized grains. This prediction is occurring even though no metadynamic recrystallization data were used to determine the model coefficients. This independent prediction of the marked change in recrystallization behavior represents a major improvement of our understanding of this process and is a critical test of the model capability.

In the following sections these and other aspects of the model basis and predictions are discussed in more detail.
5.1. Deformation

The hardening term chosen in this model is based on the assumption that the mean free path of the dislocations is linked to the presence of the substructure. Indirect methods based on the comparison of the strain rate dependence of the hardening term and the stress tended to confirm these assumptions in steel and lead [13,28]. The establishment of a substructure at the early stage of deformation has been observed in nickel [29], stainless steel [30,31] and copper [32] and depends on the recovery events taking place, i.e. the competition between hardening and annihilation. Other work using zirconium also showed that subgrain size was linked to the mean free path of dislocations [33]. Although the Estrin and Mecking model gave a relatively good fit for the work hardening rate, it has some limitations due to its simple approach, as shown by the deviation of the model from the experiments at higher Zener–Hollomon parameters (i.e. higher strain rate and lower temperature). These limitations are possibly due to the change in the establishment of substructure at higher Z, which is linked to changes in the mechanism of dislocation motion.

Despite those limitations, the model emphasizes the important contribution of recovery to the recrystallization process during and after deformation. Recovery during deformation delays the occurrence of recrystallization. Physically it also contributes to the development of the substructure necessary for the formation of nuclei. McQueen [34] showed that the substructure observed after dynamic recovery or dynamic recrystallization was very similar, indicating that dynamic recovery was important in the development of dynamic recrystallization.

Recovery is also important in the model after deformation. When unloading after a small deformation, recovery is found to be the main softening event (Fig. 2a), as observed experimentally when the level of deformation is below the critical strain for static recrystallization. When unloading after larger deformations, recovery takes place concurrently with recrystallization, although recrystallization remains the main softening mechanism. An assumption commonly used in relating softening measurements and recrystallization is that the first 10–20% of softening is due recovery appears to be valid. Recrystallization may be activated during that period but the slow growth of the nuclei does not allow static recrystallization to be observed.

The equation describing the rate of recovery is the same for both the static and dynamic processes. While it is difficult to exactly analyze the performance of the static recovery component these results do suggest that this is a valid assumption within the required accuracy limits of the model.

5.2. Recrystallization

In the present model, no critical strain was introduced to initiate recrystallization. This means that recrystallization is always activated during deformation but only becomes mathematically detectable at a certain point, referred to as the critical strain. The model also assumed no mechanistic difference between dynamic, static and metadynamic recrystallization, with the same recrystallization equation governing the three regimes. Experimentally, the distinction in mechanism between metadynamic and static recrystallization has been based on the different observed dependence of the time to a given level of recrystallization with deformation parameters such as strain rate and temperature [23,24]. The success of the current model in predicting the kinetics of static, dynamic and metadynamic recrystallization using the same basic equation suggests that the three processes are indeed similar. The apparent differences observed experimentally can be ascribed to the different ways in which the dislocation density varies with deformation conditions in the respective regimes.

The only descriptive model currently in the literature for the post dynamic process is that developed by Sakai and co-workers [35,36]. They propose that at the end of deformation, where there has been significant dynamic recrystallization, there will be three types of grains: work hardened grains with high dislocation densities, dynamic recrystallized nuclei with few dislocations and growing dynamic recrystallized grains with a dislocation density which is a function of the difference between the strain at which they are nucleated and the strain at the end of deformation. During post dynamic recrystallization they have proposed, in this model, that there is softening through growth of the dynamic nuclei, recovery of the dynamically formed grains and conventional static recovery and recrystallization of the work hardened grain. The overall softening process in this case appeared to have two stages. In the first there was true metadynamic recrystallization, but this only accounted for a small fraction of the softened fraction; most occurred by conventional static recrystallization (i.e. nucleation and growth). The current simulations would also tend to support the idea that conventional static recrystallization could cause the observed change in the sensitivity of the rate of recrystallization to the process variables as the rate of recrystallization is closely linked to the state of the work hardened grains and is relatively unaffected by the amount and level of deformation in the dynamic grains.

6. Conclusions

An internal variable model has been developed to describe the microstructures evolution during the hot
deformation of austenite. The model uses the dislocation density as the internal variable that evolves during deformation, recovery and recrystallization. A particular feature of this model is that a population distribution in the dislocation density is considered, rather than an average density.

The model considered work hardening, recovery, recrystallization and grain growth. It was found that the Estrin and Mecking model was more applicable than other established models in describing the work hardening and dynamic recovery conditions. The recrystallization part of the model was exactly the same for both static and dynamic processes.

The test data included conditions that produced dynamic recrystallization during deformation and conventional static recrystallization between deformations (i.e. there were no data for metadynamic recrystallization-static recrystallization from an initial microstructure which has dynamically recrystallized).

The model was tuned using available literature and test data to determine the material constants for a low C steel. The model was then used to extrapolate beyond the tuning data set. The important observations from this were: (a) relatively small change in the rate of static recrystallization with increasing strain for strains beyond that required to initiate dynamic recrystallization. (b) At these high strains the rate of recrystallization was largely independent of temperature, yet a strong function of strain rate.

Both of these observations are consistent with the current observations in the literature for metadynamic recrystallization, even though no data for these experimental conditions were used. The model suggested possible mechanisms to explain the previous experimental observations.

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