1. Introduction

Metallic materials deformed at elevated temperatures are subject to two basic restoration processes: recovery and recrystallization. In materials that have high stacking fault energy, such as ferrite in steels, dynamic recovery takes place, and, as a consequence, there is a reduction in the number of dislocations and sub-grain formation while the effects of work hardening are eliminated. On the other hand, in materials with low stacking fault energy, such as austenite in the steels, there is an increase in the dislocation density as the deformation increases. Eventually, local differences in dislocations density are enough to nucleate new grains during the deformation process. This nucleation during deformation promotes an annihilation of a great number of dislocations by the migration of high angle grain boundaries. This is called dynamic recrystallization (DRX).

During controlled thermomechanical processing, hot deformation of steels is carried out at temperatures located within the austenitic field of the iron–carbon equilibrium diagram. Therefore, detailed knowledge of deformation parameters and of their influence on DRX is of fundamental importance for engineering applications. This is so because, based on this knowledge, one may carefully adjust processing parameters in order to achieve a fine grained matrix. Such a fine grained microstructure, either ferritic or bainitic, yields an optimal combination of a high yield strengths and good toughness. DRX starts when a critical deformation or strain, $e_c$, is reached. Identifying this critical deformation has been the subject of several studies. Those studies show that on a flow stress vs. strain curve, $\sigma$–$\varepsilon$, the value of $e_c$ is located slightly below the peak strain. Usually the critical deformation corresponds to a stress equal to 80% of peak stress, $\sigma_f$, as illustrated in Fig. 1.
strain rate) against stress $\sigma$ of a flow curve, such as that shown in Fig. 1(b), exhibits an inflection point correspondent to $\sigma_c$. For a given strain rate, $\dot{\varepsilon}$, both the stress peak and the inflection in the $\theta$–$\sigma$ plots are attributed to initiation of DRX, see $\sigma_c$ and $\sigma_p$ in Figs. 1(a), 1(b).

For a given steel, the value of depends on the deformation temperature, strain rate and grain size, and the expression normally used to describe this dependence has the following form

$$\varepsilon_p = B D_0^p Z^{p_2} \exp \left( \frac{Q}{RT} \right)$$

and $Q$ is an activation energy.

At the end of the hot deformation processing, it may happen that austenite is deformed below the equilibrium transformation temperature. In these circumstances the deformation increases the driving force for the decomposition of this metastable austenite. This may lead to an increase in the $T_{Ar}$ temperature. If $T_{Ar}$ temperature is below the deformation temperature, austenite will dynamically transform to ferrite during deformation, see Fig. 2. $T_{Ar}$ temperature is a function of the steel chemical composition, austenite grain size, the amount of deformation and cooling rate the strain rate. This dynamic austenite to ferrite transformation is called dynamically strain induced transformation or DSIT.7–12)

Dynamically strain induced ferrite transformation has been widely studied. This interest is justified because DSIT may produce ultrafine ferrite grains, roughly 2–3 $\mu$m in size. Notice that during deformation metastable austenite may also dynamically recrystallize. The occurrence of DRX or DSIT, will depend on whether the critical deformation for DRX, $\varepsilon_{c,DRX}$, or for DSIT, $\varepsilon_{c,DSIT}$ is reached first.

Both DRX and DSIT imply in dynamic nucleation of new stress-free grains either austenitic or ferritic. These new deformation-free grains cause a reduction in the deformation stress, as shown in Fig. 1. Besides this softening, either DRX or DSIT, can also result in a final microstructure consisting of fine ferritic grains.7,9–12)

If the deformation is applied at temperatures slightly above or slightly below the austenite to ferrite equilibrium transformation temperature, $T_{Ae}$, DRX will prevail. This is so because in these circumstances $\varepsilon_{c,DRX}$ is normally less than $\varepsilon_{c,DSIT}$. By contrast, if the deformation temperature is close to $T_{Ar}$ of the steel, DSIT dominates. This is illustrated in Fig. 3 that shows the behavior of critical deformation as a function of deformation temperature for different reheating temperatures and strain rates for C–Mn steel.11)

Beladi et al.9) Dong and Sun10) and Choi et al.11) studied...
the influence of thermomechanical parameters on formation of ferrite by DSIT in C–Mn and C–Mn–V steels, with low and high Mn content. The addition of microalloying elements, Nb, Ti and V to carbon steels form nitrides and carbides during the thermomechanical processing leading to a considerable austenite grain refining. Subsequent transformation of this fine grained austenite may result in a fine grained ferrite, owing to a significant increase in the number of ferrite nucleation sites. Nb may be present either as precipitates or in solid solution, in both cases retarding grain boundary motion. Thus, Nb has a strong effect on the non-recrystallization temperature, \( T_{nr} \). An increase in Nb content results in an increase in the non-recrystallization temperature. As a result, one may deform austenite more heavily and consequently increase its stored energy during hot rolling. When Nb is in solid solution it may also retard the austenite to ferrite transformation.\(^{13}\) The effect of Nb on DSIT was studied by few researchers.\(^{13–15}\) Hong et al.\(^{13}\) did compression tests with 0.8 of true deformation in Nb steel at four temperatures, with the highest temperature being 762°C. Eghbali and Abdollah-zadeh\(^{14}\) also used compression tests with 0.80 of true deformation, temperatures of 1100, 950 and 845°C and low strain rates, 0.001 and 0.1 s\(^{-1}\). Finally, Dong and Sun\(^{15}\) developed a thermomechanical controlled rolling schedule specifically to promote DSIT. Dong and Sun applied three passes of 30% of reduction at 820°C, cooled with 20°C/s to 550°C and finally let the specimen cool slowly in a furnace to room temperature to simulate strip coiling. This processing resulted in a ferrite grain size as small as 1.5 \( \mu m \).

In a previous work, the present authors\(^{16}\) studied the effect of Nb on DSIT but used a limited range of thermomechanical parameters. This work builds on this previous work, focusing on the effect of thermomechanical parameters on the occurrence of dynamic mechanisms of grain refinement. Specifically, a C–Mn and C–Mn–Nb steel is submitted to hot torsion deformation at 900, 850 or 800°C, at strain rates of 1 s\(^{-1}\) and 5 s\(^{-1}\) and total strains of 1, 2 and 3. In addition, the degree of grain refinement that results from different mechanisms, i.e., DRX or DSIT has also been evaluated. High deformation temperatures were chosen in this work because if one could obtain DSIT at high temperatures, say above 800°C, this would be of considerable practical importance. Indeed, obtaining DSIT at high temperatures could make industrial use of DSIT viable without too much change in existing equipment.

2. Experimental

Two steels with base composition within the range of SAE 1010 specification were employed: one with and another without Nb.

2.1. Hot Torsion Test

Figure 4 describes the thermomechanical processing. Cylindrical torsion specimens were machined from two hot rolled plates. The dimensions were 50 mm in gage length and 8 mm in diameter. Figure 4 describes a sequential scheme of the experimental procedure. The thermomechanical treatment employed in this work (Fig. 4) consisted of three stages: a) soaking at 1 200°C; b) hot torsion deformation, and c) an accelerated cooling. Samples were first austenitized at 1 200°C for 60 s. At this temperature all carbonitrides were dissolved. Following this austenitization one pass of 0.5 true strain at a strain rate of 5 s\(^{-1}\) was applied to the specimen at 1 100°C. The specimen sample was then cooled with a cooling rate of 10°C/s to the desired temperature: 900, 850 or 800°C and immediately subjected to deformation. The deformation consisted of only one pass of 1.5 or 2.5 or 3.5 true strain at a strain rate for 5 s\(^{-1}\). During hot torsion the specimens were protected by a nitrogen atmosphere which helped to prevent adiabatic heating. The torsion equipment employed in the experiments was a Gleeble thermomechanical simulator, details are given elsewhere.\(^{16}\) After deformation, the group of samples was water quenched to room temperature. Stress–strain curves were determined from the torque-twist behavior in the usual manner.\(^{17,18}\)

2.2. Metallographic Characterization

Optical and scanning electron metallography was carried out using conventional methods. Longitudinal cross sections of the 0.5 mm form the surface of the torsion specimen were examined and 4% nital etching was used for the analysis. Nital is generally well adapted to reveal ferrite and pearlite in carbon and low alloy steels. However, metallography becomes more complex when the microstructure simultaneously contains austenite and martensite as second phases. The relative statistical errors in grain size of prior austenite or ferrite determination were about \pm 10\%.

The \( A_r \) temperatures were determined by dilatometry, after soaking at 900°C and cooling rate of 10°C/s. It was found 785°C for C–Mn and 787°C for C–Mn–Nb. The equilibrium temperatures \( A_e \) for C–Mn and C–Mn–Nb steels were calculated by ‘Thermo-Calc’ software, 857°C and 861°C, respectively.

3. Results

3.1. Effect of Strain Rate and Deformation Temperature

Figures 5 and 6 show the stress–strain curves for C–Mn and C–Mn–Nb steels for different deformation conditions. The resulting microstructures are illustrated in Figs. 7 and 8. These figures show the effect of temperature and defor-
Information on critical stress and deformation on the mechanism of grain refining as described below.

Figures 5(a), 5(b) show that $\sigma_p$ and its corresponding $r_p$, indicated by arrows, are higher for C–Mn–Nb. These values are also higher when the strain rate increases up to 5 s⁻¹. Figs. 6(a), 6(b). These results signify that a softening mechanism occurs during the deformation process. The first possibility is DRX of austenite. In addition, Figs. 5 and 6 also

Fig. 5. Stress–strain curves of C–Mn (a) and C–Mn–Nb (b) steel for different deformation temperatures deformed with a strain rate of 1 s⁻¹.

Fig. 6. Stress–strain curves of C–Mn (a) and C–Mn–Nb (b) steel for different deformation temperatures deformed with a strain rate of 5 s⁻¹.

Fig. 7. Micrographs of C–Mn steel deformed with strain rate of 1 s⁻¹ and 5 s⁻¹, and then water quenching. Nital 4% etched.
show that as the deformation continues and the strain increases, the $\sigma$–\(e\) curves exhibit a transition from work hardening to work hardening plus dynamic recovery (DR) and DRX or another softening mechanism. In this case, the ferrite formation from work hardening austenite takes place, mainly for lower temperatures, 850 and 800°C.

Figures 7 and 8 show that for C–Mn steel deformed at 800°C, both with 1 s\(^{-1}\) and 5 s\(^{-1}\), the microstructure consisted in ferrite and some acicular constituents, whereas for 850 and 900°C there is ferrite and a large amount of martensite. This means that at 850 and 900°C the C–Mn steel was deformed in the austenitic field.

In contrast, for the C–Mn–Nb steel deformed at 800°C and 850°C, both with 1 s\(^{-1}\) and 5 s\(^{-1}\), ferrite predominated in the microstructure. After deformation at 900°C, martensite predominated when the strain rate was 1 s\(^{-1}\) (Fig. 8(c)). But for a strain rate of 5 s\(^{-1}\), martensite was the main constituent with some grain boundary ferrite (Fig. 8(f)).

Analyzing the flow curves and the micrographs, it is possible to follow the evolution of the microstructure as function of deformation temperature. It is clear that the grain refining mechanism strongly depends on the deformation temperature.

Considering the stress peaks after the second pass (Figs. 6 and 7) the critical strain decreases as the deformation temperature increases for the both strain rates used. The correspondent peak deformations increased as the deformation temperature was lowered, from 900 to 850°C, for both steels and both strain rates. For the lowest deformation temperature, 800°C, the deformation corresponding to the peak stress decreased in the C–Mn steel, while for C–Mn–Nb this deformation did not significantly changed, as illustrated in Fig. 9.

DSIT occurred for C–Mn deformed at 800°C and C–Mn–Nb steel deformed at 800°C and 850°C. This can be inferred from to ferrite formation in the samples quenched soon after deformation ended. For deformation temperatures of 850°C and 900°C the C–Mn steel and at 900°C for C–Mn–Nb steel, the grain refinement mechanism was DRX. The equiaxed austenite grains after deformation observed at these temperatures is a strong indication of the occurrence of DRX.

It is likely that the grain boundary ferrite, observed in the sample of C–Mn steel deformed at 900°C at 1 s\(^{-1}\) (Fig. 8(c)), formed during quenching from this temperature, in which there were very fine recrystallized austenite grains, which reduced the hardenability of the steel.

### 3.2. Effect of Amount of Deformation at 800°C and Strain Rate 5 s\(^{-1}\)

The volume fraction of DSIT ferrite is directly proportional to the amount of deformation. Figures 10 and 11 show the stress–strain curves and correspondent microstruc-
tures after different level of deformation at 800°C and strain rate of 5 s\(^{-1}\). In none of the steels it was possible to obtain only ferrite (Figs. 10(c) and 11(c)). Acicular products and a small amount of pearlite can also be seen in the microstructure.

To obtain ultrafine grains by DSIT it is necessary not only to reach the critical deformation for DSIT to happen, but also to go beyond this in order to generate a significant amount of ferrite grains.\(^9,19\) As can be observed in Figs. 10 and 11, there is an increase in ferrite volume fraction when the deformation increases from 1.5 to 3.5. The remaining austenite transforms to acicular products after quenching, see Fig. 11(c). For the C–Mn steel, the ferrite volume fraction increased but its grain size also increased when compared to the C–Mn–Nb steel. However, up to the maximum total strain employed 3.5, the ferrite volume fraction continuously increased and its grain size was reduced (Fig. 10(c)). Similarly, for C–Mn–Nb steel, Fig. 11, the ferrite volume fraction increased with deformation and its grain size was

Fig. 10. Stress–strain curves and respective microstructure for different level of deformation for C–Mn steel. Etched with Nital 4%.

Fig. 11. Stress–strain curves and respective microstructure for different level of deformation for C–Mn–Nb steels. Etched with Nital 4%.
also continuously reduced.

**4. Discussion**

4.1. Influence of Thermomechanical Parameters on the Formation of DSIT Ferrite

For a given steel chemical composition, the formation of DSIT ferrite is controlled by the thermomechanical parameters. In what follows the influence of important variables, such as, prior austenite grain size, strain, strain rate and deformation temperature on the formation of DSIT are discussed in what follows.

In the present work, both steels had the same chemical composition except that one of the steels contained Nb. After the first deformation step at high temperature, 1 100°C, the austenite grain size was equal to 35 μm and 30 μm for the C–Mn and for the C–Mn–Nb steel, respectively. During cooling from 1 100°C to the temperature of the next deformation pass, 800°C, the austenitic grain sizes increased from 35 to 60 μm and from 30 to 40 μm for the C–Mn and for the C–Mn–Nb steel, respectively. This smaller prior austenitic grain size of the C–Mn–Nb steel could favor DSIT ferrite because of the larger number of nucleation sites available at the grain boundaries.

The analysis of the stress–strain curves permits to follow the microstructural evolution as a function of deformation temperature and strain rate. Previous works focusing on the above parameters are available.9–14) There are, however, significant differences between those and the present work, such as, the actual temperature used in the experiments, the amount of total deformation and the use of Nb as a microalloying element.

Figure 8(b) shows that the C–Mn–Nb steel specimen deformed with a strain rate of 1 s\(^{-1}\) either at 800°C or 850°C, could not stand the load and broke for total deformations in excess of 2. Two possibilities can be invoked to explain this. The first, less likely, is an instability in the temperature control of the thermomechanical simulator that could lead to a loss of ductility. The second, more probable, is hot shortness caused by DSIT ferrite formation at the austenitic grain boundaries. According to Mintz et al.,20) deformation will preferentially take place in this ferrite that will work harden causing intergranular fracture of the steel. Beladi et al.21) also report fracture on a C–Mn–V deformed at 775°C with a strain rate of 0.1 s\(^{-1}\) for a total deformation equal to 2.3. Beladi et al.9) pointed out that under these conditions DSIT ferrite forms preferentially at the austenitic grain boundaries.

The deformation corresponding to the peak stress, the ’peak deformation’, is plotted in Fig. 9 as a function of deformation temperature. Figure 9 shows that for the second deformation pass the peak deformation decreases with an increase in deformation temperature for the C–Mn deformed with either 1 s\(^{-1}\) or 5 s\(^{-1}\). By contrast for the C–Mn–Nb steel the peak deformation reaches a maximum value at 850°C. The lower value of the peak deformation for the C–Mn–Nb at 900°C is related with dynamic recrystallization of austenite, as shown by several authors.4–6) The lower value of peak deformation at 800°C has an entirely different reason: it is caused by the formation of DSIT ferrite, notice that 800°C is between Ar\(_3\) and Ae\(_3\). The larger is the amount of total deformation the larger is the volume fraction of DSIT ferrite formed. Returning to the C–Mn steel, one observes in Fig. 9 that the increase in the value of peak deformation does not follow a straight line. If the straight line from lower temperatures is prolonged, see dashed lines in Fig. 9, the peak deformation would have reached 1.0 and 1.2 for the strain rates of 1 s\(^{-1}\) and 5 s\(^{-1}\), respectively. This means that in the C–Mn steel DSIT ferrite is also taking place at 800°C, thus causing a reduction in peak deformation but in this case a lower amount of DSIT ferrite forms compared with the C–Mn–Nb steel. Figures 10 and 11 show the microstructures C–Mn and C–Mn–Nb steels were significantly different for the same deformation temperature and strain rate. Figures 5 and 6 show the same relationships discussed above directly in the stress strain curves, namely, that an increase in strain rate caused an increase in both peak deformation and peak stress.

4.2. Ferritic Grain Refinement by DSIT

In order to form ferrite by DSIT the deformation temperature must be located between Ar\(_3\) and Ae\(_3\). As mentioned above, Ar\(_3\) is equal to 785°C for the C–Mn and 787°C for the C–Mn–Nb steel and Ae\(_3\) is equal to 857°C for the C–Mn and 861°C for the C–Mn–Nb steel, respectively.

Table 2 shows the ferritic grain size after quenching. Notice that no ferrite formed either by DSIT or during quench for the C–Mn steel deformed at 850°C, for this reason Table 2 does not show any value for this condition. It is clear that the Nb steel had a finer ferritic grain size. An in-

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**Table 1.** Chemical composition (wt%) for the C–Mn and C–Mn–Nb steels (SAE 1010 steel specification is also shown.).

<table>
<thead>
<tr>
<th>Identification</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>P</th>
<th>S</th>
<th>Al</th>
<th>N</th>
<th>Nb</th>
</tr>
</thead>
<tbody>
<tr>
<td>C–Mn</td>
<td>0.10</td>
<td>0.50</td>
<td>0.15</td>
<td>0.017</td>
<td>0.027</td>
<td>0.030</td>
<td>0.0065</td>
<td>-</td>
</tr>
<tr>
<td>C–Mn–Nb</td>
<td>0.094</td>
<td>0.50</td>
<td>0.15</td>
<td>0.019</td>
<td>0.023</td>
<td>0.029</td>
<td>0.0064</td>
<td>0.034</td>
</tr>
<tr>
<td>SAE 1010</td>
<td>0.08–</td>
<td>0.3–0.6</td>
<td>0.15–0.35</td>
<td>&lt;0.040</td>
<td>&lt;0.050</td>
<td>-</td>
<td>-</td>
<td></td>
</tr>
</tbody>
</table>

**Table 2.** Mean ferrite grain size, \(\mu m\).

<table>
<thead>
<tr>
<th>Steel</th>
<th>Temperature, °C</th>
<th>Strain Rate (s^{-1})</th>
<th>C–Mn</th>
<th>C–Mn–Nb</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
<td>C–Mn</td>
<td>C–Mn–Nb</td>
</tr>
<tr>
<td></td>
<td></td>
<td>1</td>
<td>5</td>
<td>1</td>
</tr>
<tr>
<td></td>
<td></td>
<td>800</td>
<td>3.8</td>
<td>3.8</td>
</tr>
<tr>
<td></td>
<td></td>
<td>850</td>
<td>*</td>
<td>*</td>
</tr>
<tr>
<td></td>
<td></td>
<td>900</td>
<td>*</td>
<td>*</td>
</tr>
</tbody>
</table>

* Microstructures predominantly composed by martensite.
teresting point is that the ferritic grain size remained approximately the same for the two strain rates used here. Previous researchers generally agree that an increase in total deformation results in a decrease in ferritic grain size. When austenite is deformed at lower temperatures its stored energy increases. This higher stored energy may lead to an increase in the number of sites favorable for ferrite nucleation. Thus, lower deformation temperatures are likely to give rise to finer ferritic grain sizes. In contrast when deformation is carried out at higher temperatures, close but still below to \( \text{Ae}_3 \), a large amount of deformation will be necessary to reach the critical deformation for ferrite DSIT to occur. These conditions are shown schematically in Fig. 12. If Nb is present in the austenite, it may retard austenite recrystallization, i.e., increase the critical deformation for DRX, in this way allowing the critical deformation for ferrite DSIT to be reached first, even for a deformation temperature as high as 850°C. This is what the evidence obtained in this work suggests.

There are several other papers that report the formation of ultrafine DSIT ferrite without considering that in order for it to form a critical deformation must be reached first. As proposed by Beladi et al., for example, Choi et al. found ferritic grain 2.3 \( \mu \text{m} \) in size in a C–Mn steel for a total deformation equal to 1.2 at 750°C. We agree with Beladi et al. and present work supports the view that there is a critical deformation for the onset of DSIT ferrite. A high volume fraction of DSIT ferrite and the possibility of preventing its growth during quenching are key factors to obtain a fine ferritic grain size at the end of thermomechanical processing. Although the onset of DSIT depends on a critical deformation, the volume fraction of DSIT ferrite for a given prior austenitic grain size, will be a function of both the deformation temperature and the amount of total deformation as shown in Fig. 12. The upper part of Fig. 12 depicts the volume fraction of ferrite, its value increases from right to left. Therefore if the austenite to ferrite transformation takes place close but still below to \( \text{Ae}_3 \), the amount of deformation required for DSIT will be large and the volume fraction of DSIT ferrite will be small. In addition because ferrite forms at such a high temperature it will be very difficult to prevent some grain growth. On the other hand, for deformation temperatures well below \( \text{Ae}_3 \), the critical deformation for DSIT will be smaller than that for high temperatures and the ferrite volume fraction will be higher. Deformation at lower temperatures will then allow the formation of a large volume fraction of DSIT ferrite without too much concern for the critical deformation because at low temperatures one expects that the critical deformation for DRX will invariably be much larger than that for DSIT.

Therefore, as was also shown by Hong et al., the undercooling of austenite is relatively more important for DSIT than the applied deformation. Hong et al. increased the cooling rate after austenite deformation in a C–Mn steel to decrease \( \text{Ar}_3 \), by this method, they were able to obtain almost the equilibrium volume fraction of ferrite by DSIT in very low temperature, 680°C, with small reductions, about 30% in compression, in agreement with Fig. 12.

Another important point is the significant growth of ferrite in C–Mn steels formed by DSIT, from \( \varepsilon = 1 \) to \( \varepsilon = 1.5 \) (Figs. 10(a), 10(b)). Huang et al. remarks that DSIT ferrite, being formed during deformation, has time to undergo growth. Figure 10(c) shows that the microstructure is more uniform when \( \varepsilon = 3.5 \), because for such high deformation ferrite nucleates not only at the grain boundary but also at sites within the deformed grains. This increase in volume fraction and contact between ferrite grains apparently makes ferrite growth more difficult. In order to clarify this one must distinguish between two mechanisms by which DSIT ferrite grains may grow. The first is simply growth by consuming the surrounding austenite, a very fast process, the second is classical grain growth, in which larger ferrite grains grow at the expense of the smaller ones. It is difficult to tell which mechanism is more important in practice. Nevertheless the fact that when a high volume fraction of DSIT ferrite forms smaller increases in grain size are found suggests that classical grain growth, though also fast at high temperatures, is not as fast as ferrite growth at the expense of surrounding austenite.

It is worthy of note that in a recent work, Pan et al. investigated a series of C–Mn steels and C–Mn steels microalloyed with Nb and Nb–Ti. These steels were processed not only by thermomechanical treatments but also by cold rolling an initially martensitic microstructure in order to obtain an ultrafine grain structure. Pan et al. showed that a using a deformation by compression in the range of 0.9–1.2 at 780°C, after austenitizing at 900°C, followed by cooling at 10°C/s resulted in a final ferritic grain size in the range of 2–3 \( \mu \text{m} \). Their grain sizes were of the same order of magnitude as those obtained from the present work even though they followed a different processing route.

The problem of ferrite growth after DSIT can also be seen in the work of Hodgson et al. The authors used a hot torsion test interrupted at certain deformations in order to show an increase in the grain size of DSIT ferrite in a AISI 1010 steel deformed 1.5 at 760°C with a strain rate of 3 s\(^{-1}\). They showed that a 10 s holding time at 760°C is enough to increase the ferritic grain size from 2 to 4.1 \( \mu \text{m} \). This work shows quite clearly that any benefit derived from using a large total deformation and high strain rates can be lost if this growth cannot be prevented.

Finally, one can summarize the above discussion by saying that thermomechanical parameters do have an important influence on the formation and grain size of DSIT ferrite but all this processing effort may be lost if one is unable...
to prevent the DSIT ferrite from growing soon after its formation. This key issue, which was also observed in previous work from several authors, must surely be addressed in forthcoming papers if one is to establish a viable route for obtaining an ultrafine ferritic microstructure using DSIT.

5. Conclusions

For C–Mn steel deformed at 800°C, DSIT occurred. On the other hand for deformation temperatures of 850°C and 900°C, dynamic recrystallization of austenite (DRX) took place.

For C–Mn–Nb steel DSIT was the mechanism of ferrite formation at the deformation temperature of 800°C. In addition, Nb retarded the occurrence of DRX at 850°C allowing DSIT to take place. On the other hand, at 900°C, DRX was suppressed by Nb but DSIT could not occur because this temperature is above the austenite to ferrite equilibrium temperature.

The variations in the thermomechanical parameters: deformation temperature, strain rate and amount of deformation were responsible for changes in critical strain for DRX and DSIT. These parameters also influence the amount of ferrite that was formed by DSIT.

The mean ferrite grain sizes after DSIT and quenching for C–Mn deformed at 800°C under strain rate of 1 s⁻¹ and for 5 s⁻¹ it was 3.8 μm. For C–Mn–Nb steel deformed at 800°C, the ferrite grain size was 2.2 μm, under a strain rate of 1 s⁻¹ and 2.3 μm, for a strain rate 5 s⁻¹. After deformation of the C–Mn–Nb steel 850°C, for strain rates of 1 s⁻¹ or 5 s⁻¹ ferrite grain size of 2.9 μm and 2.8 μm, respectively, were observed.

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REFERENCES