Static Recrystallization Behaviour of a Wide Range of Austenite Grain Sizes in Microalloyed Steels

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The static recrystallization kinetics of four steels microalloyed with Nb, Ti and Nb–Ti have been investigated, covering conventional and hot direct rolling conditions. The influence of the grain size on the recrystallization time has been studied for a wide range of grain sizes (20–1000 μm). For the same deformation conditions it has been observed that the quadratic dependence on grain size works well in the range of finer grain sizes (<160 μm). In contrast, the recrystallization times measured at coarse grain sizes are significantly lower than those predicted by this dependence. However, when the whole grain size range analysed is considered, a linear dependence on grain size gives the best fit to experimental data. The effect of strain on the recrystallization kinetics has been also analysed for different initial grain sizes. It has been observed that the time to reach a 50% recrystallized fraction (t0.5) shows a dependence on strain of the type t0.5 ≈ ε−p, the exponent being influenced by the initial grain size. A unique equation has been proposed to describe the static recrystallization kinetics of Nb, Ti and Nb–Ti microalloyed steels. The predictions of the equations have been compared to previously published data on similar steels.

KEY WORDS: static recrystallization; hot direct rolling; Nb; Ti and Nb–Ti microalloyed steels.

1. Introduction

Static recrystallization kinetics in C–Mn and microalloyed steels have been investigated extensively1–19 and several regression equations have been proposed. The influence of prior austenite grain size (D0), applied equivalent strain (ε), deformation temperature (T) and strain-rate (ė) on the kinetics of static recrystallization are conveniently expressed in terms of the time required to induce some specified recrystallization fraction, say X=0.5. Experimental experience dictates that t0.5 follows an empirical relationship of the form:

\[ t_{0.5} = AD_0^m e^{-p\dot{\varepsilon}^{-q}} \exp \left( \frac{Q}{RT} \right) \] ...................(1)

where m, p and q are constants and Q is the activation energy for static recrystallization following hot deformation. Knowing the evolution of t0.5 with the thermomechanical variables and the Avrami exponent, n, the static recrystallization behaviour can be expressed by the modified Avrami equation:

\[ X = 1 - \exp \left( -0.693 \left( \frac{t}{t_{0.5}} \right)^{\frac{n}{m}} \right) \] ...................(2)

Much of the work has been devoted to various aspects of static recrystallization, notably in the range corresponding to conventional hot working processes. In consequence, most models developed to predict the static recrystallization kinetics of austenite1–19 do not take into account a wide range of austenite grain sizes, but only consider a narrower spectrum.

It is worth emphasising that there is a marked difference between the static recrystallization equations proposed for similar steels. In particular, the power exponent of strain varies from −4 to near −2.71 The exponent m of D0 takes a value between 1.7–9 and 2.15,10 Different exponents lead to significant differences in predictions. For example, the extrapolation of the inadequate coefficient, obtained in the lower grain size range, to the large grain size (typical of hot direct rolling processes) would result in a significant error.

The new trend in steel processing technology is to link the continuous casting and hot rolling operations with attendant economic advantages.8,14–19 “Thin slab casting”, “Hot direct rolling” or “Hot charge rolling” are examples of the application of this technology. However, there are significant metallurgical differences between these technologies and the conventional rolling process. The as-cast austenite is characterised by a very large grain size (600–1400 μm),14 in contrast to the as-reheated austenite where the grain size is significantly refined by the phase transformations on cooling and reheating.39 For microalloyed steels, the as-cast austenite is highly supersaturated with microalloying elements when compared to as-reheated austenite, which would be expected to affect precipitation phenomena. However, the mode and kinetics of recovery and recrystallization of these coarse grains have not been studied extensively. Similarly, the interactions between deformation, precipitation and recrystallization in the as-cast
austenite are not yet completely understood. These differences require an extension of the valid intervals of application for the equations already obtained. The slower recrystallization of the very coarse austenite in C–Mn and microalloyed steels in hot direct rolling processes, allied to the high supersaturation of microalloying elements, can modify the exponent values of Eqs. (1) and (2).

A study has been carried out to determine the effect of initial austenite grain size on the 50 % recrystallization time, by using a wide range of grain sizes (20–1 000 μm) in one Nb, two Ti and one Nb–Ti low carbon microalloyed steels. The influence of the applied strain on the static recrystallization kinetics of these microstructures has also been investigated.

2. Material and Experimental Procedure

The chemical compositions of the steels analysed in wt% are listed in Table 1. Three grades of steel have been considered. The first steel is a Nb microalloyed steel (Nb steel). The second steel (Nb–Ti) has a similar Nb content to the first, but also contains 0.067 wt% of Ti. The other two steels (Ti-a and Ti-b) have Ti additions with different N levels, their Cu contents being similar to those steels produced in an electric arc furnace with scrap.

In order to study the static recrystallization kinetics, two-pass torsion tests were carried out, under isothermal conditions, to examine the softening behaviour of the austenite between intervals of hot working. The geometry of torsion specimens was a reduced central gauge section, 15.5 mm in length and 7.5 mm in diameter. The specimens are reheated at different soaking temperatures in the range from 1 100 to 1 460°C for 12–15 min. and cooled to test temperature (1 100 and 1 150°C). Once the temperature is stabilised, the specimen is prestrained at a constant strain rate of 1 s⁻¹, unloaded and held for increasing times. After the interruption, the specimen is reloaded at the same deformation condition and temperature. The heating cycles were performed with induction equipment and during the tests specimen oxidation was prevented with an argon atmosphere. Coarse austenite microstructures have been produced by reheating at very high temperatures in the range from 1 400 to 1 460°C.

The initial austenite grain size was determined from samples quenched after the soaking treatment. The microstructure was revealed using an aqueous picric acid etchant with a wetting reagent. The measurements of grain size were performed using a quantitative image analyser.

The study of the recrystallization kinetics has been focused on the influence of the initial austenite grain size and the applied deformation. Grain sizes ranging from 23 to 1 073 μm and pass-strains between 0.1 and 0.6 have been considered. The fractional softening has been determined by the 2 % offset method, which excludes the effect of recovery better than the 0.2 % offset method.¹ The details of this technique have been described elsewhere.¹²⁰ In a recent work, a good correlation has been observed between the fractional softening quantified by this method and the metallographically measured recrystallized fraction.²⁰ On the other hand, for some specific deformation conditions, single-pass torsion tests were conducted, and the static recrystallized fraction was metallographically measured and compared with the softening quantified by mechanical procedure (two-pass torsion tests).

3. Results

Figure 1 shows the evolution of austenite grain size with reheating temperature for the steels analysed in the present work. From the figure, it can be observed that grain coarsening occurs at the lowest temperature for the Nb steel and at the highest for the Ti-a steel, the Nb–Ti and Ti-b steels having an intermediate behaviour.

Figure 2 shows the variation of the softening fraction as a function of time for the case of the Nb steel reheated at

<table>
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<th>Steel</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>S</th>
<th>P</th>
<th>Cr</th>
<th>Ni</th>
<th>Cu</th>
<th>Al</th>
<th>Nb</th>
<th>Ti</th>
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<tr>
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<td>0.31</td>
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<td>0.018</td>
<td>&lt;0.02</td>
<td>&lt;0.02</td>
<td>&lt;0.02</td>
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<td>Nb-Ti</td>
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<td>0.006</td>
<td>0.011</td>
<td>0.02</td>
<td>0.02</td>
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<td>0.053</td>
<td>0.034</td>
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<tr>
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<td>0.25</td>
<td>0.05</td>
<td>0.006</td>
<td>0.007</td>
<td>0.07</td>
<td>0.07</td>
<td>0.21</td>
<td>0.034</td>
<td>-</td>
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<td>0.0079</td>
</tr>
<tr>
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<td>0.042</td>
<td>-</td>
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<td>0.0037</td>
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Fig. 1. Grain growth of the microalloyed austenite as a function of the reheating temperature. The 95 % confidence limits are included.

Fig. 2. Influence of the initial grain size on softening in the Nb steel (ε=0.3, T_{heating}=1100°C).
different temperatures, when varying the initial austenite grain size from 65 to 806 μm. A strain per pass of ε=0.3 and a deformation temperature of 1100°C have been used in these tests. Similarly, the evolution of the softening fraction with time for the case of the Nb–Ti steel after reheating at different temperatures is plotted in Fig. 3. In the latter tests the strain per pass and the deformation temperature are ε=0.2 and 1100°C, respectively. Similar tests were also carried out on the Ti steels, using in this case a deformation temperature of 1150°C and a strain per pass of ε=0.3, and the results are illustrated in Fig. 4. From Figs. 2 to 4 it can be clearly observed that in general the softening curve moves towards longer times as the reheating temperature increases (larger grain size).

On the other hand, it is worth emphasising that the influence of grain size on the static softening delay is not the same in all the analysed range. For example, in the Nb–Ti steel (see Fig. 3) while for small grain sizes there is a significant delay in the static softening kinetics (from $D_0=60$ to 109 μm, for instance), in the coarser grain size interval, this effect is notably smaller (for example from $D_0=392$ to $797$ μm). A similar behaviour can be deduced from the softening data obtained with the other analysed steels.

Figures 5 and 6 show the effect of the strain on the static softening curves obtained for the Nb and Nb–Ti microalloyed steels, respectively. Figure 5 corresponds to the Nb steel reheated at 1400°C, in this case the austenite grain size being 806 μm. The data of Fig. 6 corresponds to the data obtained for the Nb–Ti steel reheated at 1100, 1200

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**Fig. 3.** Influence of the initial grain size on softening in the Nb–Ti steel ($T_{def}=1100°C$, $\varepsilon=0.2$, $\dot{\varepsilon}=1s^{-1}$).

**Fig. 4.** Influence of the initial grain size on softening in the Ti steels ($T_{def}=1150°C$, $\varepsilon=0.3$, $\dot{\varepsilon}=1s^{-1}$). The open symbols correspond to Ti-a steel and the closed symbols to Ti-b steel.

**Fig. 5.** Effect of strain on softening in the Nb steel deformed at $T_{def}=1100°C$ and previously reheated at 1400°C ($D_0=806$ μm). For comparison, metallographic measurements of recrystallized fraction are also included.

**Fig. 6.** Effect of the strain on softening in the Nb–Ti steel deformed at $T_{def}=1100°C$ and previously reheated at (a) 1400°C ($D_0=60$ μm); (b) 1200°C ($D_0=109$ μm); (c) 1420°C ($D_0=797$ μm).
and 1420°C, which lead to initial austenite grain sizes of 60, 109 and 797 μm, respectively. For each grain size the effect of the applied deformation on the static softening has been studied. All these tests were performed using the same deformation temperature, 100°C, and pass-strains ranging from 0.1 to 0.6. In Fig. 5 and for the case of a strain of 0.3, the metallographically measured recrystallized fractions, obtained when single-pass torsion tests are applied and after a predefined holding time the specimen is water-quenched, are also included.

The data of Figs. 5 and 6 indicate that an increase in the strain leads to an acceleration of the softening kinetics. It can be also observed that the softening curves obtained for the largest grain size are similar at strains of 0.5 and 0.6 (see Fig. 6(c)), denoting that at certain levels of deformation its variation has a small effect. On the other hand, in the same figure it can be observed that for the low strain level of 0.1, the softening has not been completed for this coarse grain size (797 μm). However, the softening has been completed for the same low deformation and steel, when lower initial grain sizes (60 and 109 μm) are employed (Figs. 6(a), 6(b) and Fig. 7).

4. Discussion

4.1. Grain Coarsening

For all the steels studied, the austenite grain growth is severely inhibited until the grain coarsening temperature. The differences observed in the austenite grain growth between the steels can be related to the nature of the precipitates present in each case. In the Nb steel the dissolution of the particles, which exert the pinning effect on grain boundaries and prevent the coarsening, occurs at lower temperatures than in the Nb–Ti and Ti microalloyed steels. In the latter case, it is well known that the presence of the very stable TiN particles can withstand dissolution at higher temperatures than those of Nb(C,N), maintaining the fine grain austenite microstructure at higher temperatures.

An abrupt abnormal grain growth typical of Nb microalloyed steels is produced near 1200°C for the Nb steel. For a similar Nb content, the grain coarsening temperature increases and the austenite grain size decreases when Ti is added. The addition of Ti also affects the dissolution of Nb, due to the fact that part of the Nb will be tied up as (Ti,Nb)N. For the same temperature, the amount of the undissolved Nb increases with Ti content. In consequence, it can be deduced that the high Ti concentration present in the Nb–Ti steel exerts a very significant influence on the grain coarsening behaviour.

As was pointed out previously, the low solubility of TiN particles leads to a higher grain coarsening temperature in the Ti microalloyed steels. The different behaviour observed between the low and high Ti microalloyed steels can be related to the different Ti/N relationships. It is well established that hyperstoichiometric Ti/N ratios tend to promote the formation of precipitates at very high temperatures, approaching the liquid state, which are very coarse and stable. These coarse particles are much less effective in pinning grain boundaries. This could explain why the Ti–C steel (Ti/C=6.5), and also the Nb–Ti steel (Ti/N=15.6), exhibit grain coarsening at lower temperatures than the Ti–a steel (Ti/N=2). For hypostoichiometric Ti/N ratios, the TiN precipitation fraction formed during solidification will be smaller, and combined with the lower Ti supersaturation level, there will be an increased tendency to form smaller particles that are much more effective in grain boundary pinning.

4.2. Effect of Initial Grain Size and Strain on Recrystallization Kinetics

In Figs. 2 to 7 two different behaviours can be observed: most of the curves have sigmoidal shapes (continuous curves), but in Fig. 6(c) and Fig. 7 the discontinuous curve shows a plateau being the softening not completed for the range of long times used.

In a previous work, it was demonstrated that in the case of plain sigmoidal softening curves the fractional softening determined by the 2% offset method was linearly related to the volume fraction of static recrystallized grains, being the softening due to recovery adequately excluded. The good correlation between the recrystallized and softening fractions determined by this procedure can be observed in Fig. 5, in which some metallographic measurements are included for comparison.

Similar single-stage softening curves as that presented in Fig. 6(c) and Fig. 7 by the discontinuous line, have been obtained by several authors. They suggested that these types of curves saturated at certain softening fraction are obtained when the static recovery is the unique softening mechanism operating. Thus, it has been supposed that the applied strain ε=0.1 is below the critical strain necessary for the occurrence of static recrystallization and therefore all the softening achieved in this condition would be due to static recovery. This argument is corroborated by the fact that the observed softening is too much delayed to be attributed to static recrystallization, according to the kinetics observed at higher strains. If the static recrystallization mechanism is activated, the deformed material will be rapidly consumed, avoiding in these regions further softening by recovery. In absence of recrystallization, the degree of softening reached only by recovery can be higher than that attained when both mechanisms act competitively.

In all the conditions investigated except for the single-stage softening curve of Fig. 6(c) and Fig. 7, the softening fraction will be considered mainly due to the static recrys-
tallization. Therefore, in the present work it will be assumed that the softening fraction represents the static recrystallization fraction. The continuous curves in Figs. 2 to 7 correspond to the fit of experimental data to Avrami equations of type (2). The static recrystallization kinetics for the Nb steel (Figs. 2 and 5) and for the two Ti steels (Fig. 4) are reasonably fitted to Avrami equations with an exponent \( n = 1 \), independent of the deformation parameters. However, in the case of the Nb–Ti steel, the Avrami exponent \( n \) is close to 0.9 in all the cases studied, except for the tests performed with the smallest grain size \( D_g = 60 \mu m \) and strains of 0.1 and 0.2. In this case, for \( \varepsilon = 0.1 \), the \( n \) exponent takes a value of 2, while for \( \varepsilon = 0.2 \) the exponent value is 1.2.

The behaviour observed for the Nb–Ti steel at the conditions of \( \varepsilon = 0.1 \) and \( D_g = 796 \mu m \) is probably due to the combination of large grain size (less grain boundary per unit area, so less nucleation site density for recrystallized grains) and low strain (less driving force for static recrystallization) preventing the occurrence of recrystallization after deformation. For the same low deformation but with a grain size of 109 \( \mu m \), the critical strain for static recrystallization is surpassed (see Fig. 6(b) or Fig. 7) and the recrystallization proceeds without problems, the Avrami exponent being \( n = 0.9 \). However, when the same low deformation (\( \varepsilon = 0.1 \)) is applied to the smallest grain size analysed for this steel (60 \( \mu m \)) the recrystallization kinetics is accelerated, resulting an Avrami exponent \( n = 2 \), as can be observed in the curves of Fig. 7.

The values of the Avrami exponent \( n \) obtained in the present work are in the range of those reported in the literature for different steels.\(^{5,9,12,25}\) Laaarsrou and Jonas\(^ {13} \) proposed a value of \( n = 1 \) for C–Mn and Nb microalloyed steels. Medina et al.\(^ {14} \) suggested a slight dependence on the temperature the obtained values of \( n \) being quite similar, between 0.62 and 1.50. In both cases, these values are very close to the results obtained in the present work in the range of coarse grain sizes or high deformation. On the other hand, values of \( n \) close to 2 have also been reported in the bibliography. For example, Sellars\(^ {5} \) and Ruibal et al.\(^ {10} \) obtained \( n = 2 \) for the case of low-alloy steels.

With respect to the effect of initial grain size on the Avrami exponent, Barraclough and Sellars\(^ {29} \) obtained a decrease from 2 to 1 in \( n \) as the initial grain size increases from 140 to 530 \( \mu m \) in a stainless steel for a strain of 0.5. However, in the present work, when the applied strain is high (0.2–0.3), there is not a significant difference in the Avrami exponent value with the grain size, as can be deduced from the curves of Figs. 2–4, the range of observation being between 23 and 1 073 \( \mu m \). Nevertheless, a different behaviour is observed in the range of small grain sizes for the Nb–Ti steel. As can be deduced from Fig. 6(a) in the case of \( D_g = 60 \mu m \), the applied strain modifies the value of \( n \). Summarising, in the majority of the conditions analysed in the present work, \( n \) is a constant and close to 0.9–1, but the value of the exponent increases up to 2, for some specific combinations of small grain size and applied strain.

The times for 50% recrystallized fraction (\( t_{0.5} \)) have been determined from the corresponding recrystallized fraction vs. time curves. The values obtained for the Nb and Nb–Ti steels after reheating at different temperatures are plotted against the strain on a log–log scale in Fig. 8. The reheating temperatures used for the Nb–Ti steel were 1 100, 1 200 and 1 420°C which give initial grain sizes of 60, 109 and 797 \( \mu m \), respectively. For the Nb steel a reheating temperature of 1 400°C was used, leading to an initial grain size of 806 \( \mu m \). The results shown in Fig. 8 support the validity of the empirical power law relationship \( t_{0.5} \propto \varepsilon^{\eta} \), in accordance with the dependence usually reported in the literature.\(^ {1–9} \)

It is worth emphasising that the above power function is only valid in the deformation interval corresponding to static recrystallization, that is above the critical strain necessary for recrystallization to proceed and below the critical strain for the onset of dynamic recrystallization. The minimum value for the onset of dynamic-recrystallization can be evaluated using the relationships proposed by Sellars \( \varepsilon \leq 0.8 \varepsilon_p \) or by Sun et al. \( \varepsilon \leq 5/6 \varepsilon_p \)\(^ {13} \) where \( \varepsilon_p \) is the peak strain present in a material which exhibits dynamic recrystallization. With the help of dynamic recrystallization equations,\(^ {30} \) it transpires that the deformation interval employed in the present work lies within to the static recrystallization range for all the conditions.

Most of the authors have proposed constant strain exponent values between −2 and −4. For C–Mn and Nb steels, Sellars\(^ {5} \) found a dependence of \( t_{0.5} \propto \varepsilon^{-4} \). Karjalainen et al.\(^ {11} \) determined a relationship of \( t_{0.5} \propto \varepsilon^{-3.3} \) for a Nb microalloyed steel with a grain size of 60 \( \mu m \), reporting exponent values between 3 and 4 for C–Mn steel with a fine grain size of 17 \( \mu m \). For Ti, Ti–Ni–Nb and Ti–Ni–V microalloyed steels with grain sizes in the range 40 to 56 \( \mu m \), exponent values between 2.5 and 3 were reported by Airaksinen et al.\(^ {31} \). In contrast, the variations of \( p \) observed in the present work suggest that this parameter is influenced by the initial grain size \( D_g \). For the two coarse grain sizes of \( \sim 800 \mu m \), a value of \( p = 2 \) is obtained for both steels, while for the Nb–Ti steel with lower initial grain sizes of 60 and 109 \( \mu m \) the value of \( p \) is close to 3.

From the previous data, it is worth emphasising that there is a big scatter in the values obtained for the strain exponent by different authors. In Fig. 9 the values of the strain exponent (\( p \)) obtained in the present work have been plotted against the grain size, together with several values reported in the literature.\(^ {11,12,25,26,31,32} \) From the figure, it is apparent...
that there is a tendency for a diminishing $p$ as $D_o$ increases.

On the other hand, as can be observed in the figure, the values of the present work corresponding to the range of medium grain sizes (60–109 µm), are within the scatter of most of the data reported in the literature.

Although in most of the published works $p$ is considered constant, some authors proposed an effect of the Nb content on the exponent value, while in other cases a dependence of the type $p \propto D_o^n$ has been proposed. The results obtained in the present work have been related to this latter expression and, in order to take into account a wide range of grain sizes, the data of the bibliography plotted in Fig. 9 for the smallest grain sizes ($<25$ µm) have also been included in the regression. The relationship obtained is as follows:

$$ p = 5.6D_o^{-0.15} \quad \text{.................(3)} $$

However, it has to be pointed out that in the literature $p$ values as low as 1.5 for C–Mn and in the range of 1.87–2.25 for Nb and Ti steels have also been proposed. These values are below the predictions obtained by the Eq. (3).

The dependence of $t_{0.5}$ on the initial austenite grain size is shown in Fig. 10. The discontinuous lines represent the quadratic dependence of $t_{0.5}$ on the initial grain size ($t_{0.5} \propto D_o^2$). Values close to this quadratic dependence have been often reported in the literature in the range of grain sizes obtained for conventionally reheated austenite ($D_o = 12–250$ µm), although some authors have proposed a linear relationship for the same range of low grain sizes. From the figure, it can be observed that, for the austenite grain sizes which are in this range (below 160 µm), a good agreement is obtained between the experimental data and the quadratic dependence. However, the recrystallization times obtained in the range of larger grain sizes are shorter than those predicted by the discontinuous lines. In this range a lesser dependence of $t_{0.5}$ on grain size has been obtained, as is shown by the continuous lines in Fig. 10.

In microalloyed steels, as the reheating temperature increases there is a grain growth accompanied by the dissolution of precipitates. It is well known that the solid solution of microalloying elements dissolved in the austenite can retard recrystallization, even though they are less effective than particles. Consequently, the 50% recrystallization times shown in Fig. 10 are influenced simultaneously by the initial grain size and by the solute concentration of the microalloying element. In the present tests, as the grain size increases, the amount of solute becomes greater. In consequence, a major drag effect will take place at the largest grain sizes, leading to longer recrystallization times than would be obtained if solute drag did not occur. In order to differentiate the contribution of both effects, it is necessary to quantify the amount of microalloying element dissolved at each reheating temperature and also to evaluate how this amount of solute affects the recrystallization kinetics.

Several experimental equations have been deduced to predict the influence of solutes on static recrystallization kinetics. In the case of Nb microalloyed steels, the following expression proposed by Dutta and Sellar is commonly used to quantify the solute effect:

$$ t_{0.5} \propto \exp \left[ \frac{275,000}{T} \right] \cdot [C] \quad \text{..............(4)} $$

where $[C]$ is the Nb concentration in solution in wt% and $T$ is the deformation temperature. The amount of niobium in solution can be calculated from the solubility product equation proposed by Irvine et al.

Jonas studied the solute effect of different microalloying elements on static and dynamic recrystallization. They found that Nb is the most potent element in producing solute drag and retarding static and dynamic recrystallization, followed by Ti, Mo and V, in decreasing order of effectiveness. These authors defined a solute retardation parameter for static recrystallization (SRP), describing the retardation produced by an addition of 0.1 wt% of the different alloying elements. For niobium this parameter reached a value of SRP=222, nearly three times (2.67) higher than the value of SRP=83 obtained for titanium. Using these parameters, to obtain the same drag effect as that produced by 0.1 wt% Ti, an amount of 0.0374 wt% Nb is sufficient. Following this, Eq. (4) could also be applied to Ti and Nb–Ti steels if the amount of element concentration in solution is corrected by a multiplying factor. Considering an
independent retardation effect for each element, the parameter $[C]$ of Eq. (4) can be defined as:

$$[C] = 0.374 \left[ \text{[Ti]} \right] \quad \text{for Ti steels} \quad \ldots \ldots \ldots (5)$$

$$[C] = [\text{Nb}] + 0.374 \left[ \text{[Ti]} \right] \quad \text{for Nb–Ti steels} \quad \ldots \ldots \ldots (6)$$

where $[\text{Nb}]$ and $[\text{Ti}]$ are the niobium and titanium concentrations in solution in wt%.

For the Ti steels, the amount of titanium present in solution at each reheating temperature has been calculated from the solubility product equations proposed by Matsuda and Narita for titanium carbide. In the case of the Nb–Ti steel, the niobium and titanium concentrations in solution at each reheating temperature were determined by using the regular solution model proposed by Adrian for steels containing combined microalloying additions.

From the above, the data shown in Fig. 10 were corrected by the effect of the strain using Eq. (3) and by the effect of solute drag using Eq. (4) with the corresponding concentration in solution in wt%. The results are shown in Fig. 11 (some other data obtained at different strain levels to those presented in Fig. 10 are also included), where the $t_{0.5}^{*}$ parameter is defined as:

$$t_{0.5}^{*} = \frac{t_{0.5}}{e^{-5.6D_0^{1.45}} \cdot \exp \left( \frac{275 000}{T} - 185 \right) \cdot [C]} \quad \ldots \ldots \ldots (7)$$

From the figure, it can be observed that the data corresponding to Nb and Nb–Ti steels fall into the same curve, while for Ti steels the obtained values are lower. The differences are due to the different deformation temperature employed in the tests carried out with Ti steels. It can be also observed that in the range of lower grain sizes (<160 μm), the quadratic dependence of $t_{0.5}^{*}$ on grain size continues giving a good fit to the experimental data. However, in the range of larger grain sizes the quadratic dependence leads to significant errors in recrystallization times.

The data shown in Fig. 11 for all the grain size range have been fitted by a unique dependence resulting in $t_{0.5}^{*} \propto D_0^{1.45}$. This expression allows a better fit in the wide range of grain size analysed, although it must be pointed out that the agreement is poorer in the range of lower grain sizes, compared to the quadratic dependence. A similar linear dependence was obtained by Patel et al. for V microalloyed steels, correcting the drag effect of the V and including in their analysis a coarse as-cast austenite grain size of 1150 μm.

### 4.3. Modelling of the Recrystallization Kinetics

The static recrystallization kinetics can be described by the modified Avrami Eq. (2), where $X$ is the recrystallized fraction at the time $t$, and $f_{0.5}$ represents the time required to reach a 50% recrystallization. For the present steels, it has been observed that the average value of the Avrami exponent $n$ is nearly independent of the deformation parameters and is in the range of 0.9–1 for all the steels.

The general expression for $f_{0.5}$ includes the effect of almost all the variables that intervene in hot deformation, following a law of the type expressed in Eq. (1). The effect of strain rate is less than the influence of other variables and is commonly neglected by many authors. However, according to other authors its effect is not insignificant, leading to a decrease on the recrystallization time as the strain rate increases. In the present work, the dependence of $f_{0.5} \propto e^{-0.53}$ determined by Medina et al. has been adopted.

When the solute drag effect is corrected the value of the activation energy to be considered in Eq. (1) should be one corresponding to a C–Mn steel. In the present work a value of $Q=180$ kJ/mol has been chosen. As was previously mentioned, the effect of solute concentration on static recrystallization will be taken into account by using Eq. (4). Considering all the aforementioned assumptions, the static recrystallization $f_{0.5}$ time can be represented fairly well with the following equation for the four analysed steels:

$$f_{0.5} = 9.92 \times 10^{-11} \cdot D_0^{0.53} \cdot e^{-5.6D_0^{1.45}} \cdot e^{-0.53} \cdot \exp \left( \frac{180 000}{RT} \right) \cdot \exp \left( \frac{275 000}{T} - 185 \right) \cdot \left([\text{Nb}] + 0.374[\text{Ti}]\right) \quad \ldots \ldots \ldots (8)$$

The experimental data and the values calculated using Eq. (8) have been illustrated in Fig. 12 and, as can be observed, a good correlation is obtained. Finally, some data corresponding to Nb and Ti microalloyed steels have been taken from the literature and compared with the values calculated using the Eq. (8), as illustrated in Fig. 13. For the Nb steels, the reported experimental values correspond to compositions ranging from 0.01–0.09 wt% Nb, 0.06–0.18 wt% C, 0.004–0.0119 wt% N and initial austenite grain sizes between 25 and 210 μm. The tests were performed by double deformation technique: in torsion or compression. The reported experimental values for Ti steels correspond to compositions ranging from 0.021–0.075 wt% Ti, 0.15–17 wt% C, 0.0072–0.0105 wt% N, and initial grain sizes between 29 and 95 μm. From Fig. 13 it is seen that the predictions of Eq. (8) are in reasonable agreement with the data for the times of 50% recrystallized fraction previously reported for different Nb and Ti microalloyed steels, with the majority of initial grains sizes significantly lower than those used in the present work.
dependence of tends to decrease as the grain size increases resulting in a following a dependence which can be expressed as $t_{0.5} \propto e^{-p}$. However, it has been observed that the strain power exponent $p$ depends on initial grain size. This value tends to decrease as the grain size increases resulting in a dependence of $p = 5.6D_0^{-0.15}$.

An equation for the dependence of the time for 50% recrystallization, $t_{0.5}$, on hot deformation parameters (pre-strain, strain-rate, temperature), austenite microstructure (initial grain size) and composition (solute drag effect of microalloying elements) has been derived. This equation has been shown to be generally applicable to Nb, Ti and Nb–Ti microalloyed steels to calculate the recrystallization time over a wide range of initial austenite grain sizes and deformations.

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