The Ferrite Transformation in Hot Deformed 0.036% Nb Austenite at Temperatures Above the A_{e_3}

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A 0.036% Nb microalloyed steel was deformed in torsion over the temperature range 816–896°C in a 2% H_2 –Ar gas atmosphere. Strains of 0.5–5.0 were applied at strain rates of 0.04 and 0.4 s⁻¹. The experimental parameters were varied in order to study the effects of strain, strain rate and temperature on the formation of ferrite by dynamic transformation (DT) at temperatures above the A_{e_3} . The critical strain for ferrite formation by DT was 0.5 and the volume fraction formed increased with strain and slightly with strain rate. It was also observed that the applied strain has a far greater influence on the transformation than the time. Average ferrite grain sizes of 2 to 3.5 μ m were produced, the size increasing with the transformation temperature and decreasing strain rate. By comparison with the behavior of plain C steels, it is evident that the addition of niobium slows the reverse transformation to a considerable degree. Two stages were detected in the reverse transformation: i) in stage I, observed during the initial 200 s of isothermal holding, the deformation-induced ferrite was fairly stable; ii) in stage II, observed after 200 s of holding, the reverse transformation began to take place, going to completion in about 400 s. The results of these experiments support the view that it is the stored energy of the ‘inhomogeneously’ distributed dislocation (*i.e.* those in shear bands and sub-boundaries) that provides the driving force for such “non-equilibrium” transformation.

KEY WORDS: deformation-induced ferrite; dynamic transformation; reverse transformation; A_{e_3} ; hot rolling.

1. Introduction

In pioneering work carried out in the 1980's, Yada and co-workers succeeded in transforming austenite into ferrite at temperatures above the A_{e_3} as long as the former was undergoing deformation.^{1,2)} They investigated the behavior of two low C steels containing 0.11%C–1.00%Mn–0.02%Si and 0.14%C–1.06%Mn–0.33%Si and made use of a plane strain compression machine as well as of rolling trials carried out on a two-stand experimental facility. The steels had A_{e_3} temperatures of 838 and 834°C, respectively, and had been subjected to deformation at 900°C, *i.e.* about 65°C above the equilibrium transformation temperature. They produced ferrite grain sizes of 1 to 2 μ m as long as a critical strain of about 0.5 was exceeded and observed that the ferrite volume fraction increased with deformation from about 5% at the critical strain to about 70 to 80% at accumulated strains of 3 to 4. An important feature of these results was the observation that the ferrite began re-transforming into austenite when it was held above the A_{e_3} for times exceeding about 1 s (in the first—a laboratory-melted steel) and 8 s (in the second—a commercial steel).

Despite the considerable importance of this phenomenon, there was little follow-up to the work of Yada *et al.* in the 1990's. In 2003, Chen and Chen used laser dilatometry to study the reverse transformation of ferrite in a 0.18%C–0.60%Mn–0.21%Si steel.³⁾ They observed, for example, that reverse transformation took place in their material dur-

ing about 40 s after hot compression at 860°C. In 2004, Tong *et al.*⁴⁾ used Monte Carlo methods to model the formation of ferrite from austenite above the A_{e_3} as well as its reverse transformation into austenite. This model was later extended by Xiao *et al.*,⁵⁾ who allowed for the occurrence of dynamic recrystallization (DRX) as well as dynamic transformation (DT) during straining. Their simulations indicated that the two mechanisms could take place concurrently, with DRX being favored at the higher temperatures and lower strain rates and DT under the opposite conditions.

More recently, Sun *et al.*⁶⁾ investigated the transformation behavior of a 0.17%C–0.71%Mn–0.27%Si steel for which the A_{e_3} was determined by Thermo-Calc to be 835°C. Their samples were deformed in compression at 840 and 860°C in a Gleeble thermomechanical simulator. The critical strain for the initiation of the dynamic transformation was determined using the double differentiation method, while the reverse transformation after deformation and during holding was followed by means of a quartz dilatometer. They showed that the ferrite formed during austenite deformation re-transformed into austenite at both 840 and 860°C during the first 200 s of holding after deformation.

The reason why ferrite can form above the “equilibrium transformation temperature” has been considered by a number of authors. For example, Parker⁷⁾ evaluated the extent to which the driving force for transformation increases as a re-

sult of the stored dislocation content of the deformed (and unrecrystallized) austenite. She obtained values of 10^6 – 10^7 J m^{-3} (7.2 – 72 J mol^{-1}) for the increase in the driving force. These calculations were taken somewhat further by Hanlon *et al.*,⁸⁾ who estimated the effect on A_{e3} of a ‘homogeneous’ distribution of dislocations as well as of a well-developed substructure. The presence of either of these led to a 10°C increase in the effective A_{e3} , which is much less than was observed by Yada and co-workers. Sun *et al.*⁶⁾ also calculated the effect of the stored energy on the driving force. On the basis of an assumed ‘homogeneous’ dislocation density, they obtained a free energy increase in the austenite of $3.1 \times 10^6 \text{ J m}^{-3}$ (22.3 J mol^{-1}), which they considered too low to account for the effect by a factor of about two. They thus concluded that it must be the ‘inhomogeneously’ distributed dislocations, *e.g.* those in shear bands and near grain boundaries, that are responsible for promoting nucleation above the A_{e3} . Near these ‘concentrations’ of dislocations, the driving force for transformation is increased by considerably more than the average value.

In the present investigation, the formation of ferrite above the A_{e3} (and its reverse transformation into austenite) were studied in a hot deformed microalloyed steel. Of interest was the determination of the effect of Nb addition on the kinetics of ferrite formation and reverse transformation. Unlike the previous studies, straining was carried out in a hot torsion machine. This enabled the samples to be deformed to larger strains than heretofore and also provided much more data for analysis, as the dependence of the local strain on radius produces multiple observations in a given sample. Finally, the dependence of the rate and extent of ferrite formation and austenite re-formation on temperature, strain rate and strain were evaluated by means of optical microscopy.

2. Experimental Procedure

2.1. Torsion Testing

The steel investigated in the present study was supplied as a hot rolled plate 12.5 mm thick. Cylindrical torsion specimens 3.15 mm in radius and 22.4 mm in length were machined from the plate, the cylinder axes being parallel to the rolling direction. The chemical composition and A_{e3} temperature of the steel are given in **Table 1**. The latter was determined by both the FACT⁹⁾ and Thermo-Calc thermodynamic softwares.¹⁰⁾

The Nb solution temperature was determined to be 969°C by mean of the expression¹¹⁾ below. An austenitization temperature of 1200°C was therefore selected to simulate reheating.

$$\begin{aligned} \text{Log}[\text{Nb}] \left[C + \frac{12}{14} N \right] \\ = 3.14 - 8049/T - [\text{Mn}](0.91 - 1371/T) \\ + [\text{Si}](0.35 - 923/T) \dots\dots\dots(1) \end{aligned}$$

A schematic view of the thermomechanical schedule for the deformation tests is provided in **Fig. 1**. A horizontal tubular furnace with a controlled atmosphere composed of argon and 2% H_2 was employed at all times to minimize de-

Table 1. Chemical composition (mass%) and equilibrium temperature A_{e3} of steel investigated.

C	Si	Mn	Al	Nb	P	S	$A_{e3}(^\circ\text{C})^*$	$A_{e3}(^\circ\text{C})^{**}$
0.09	0.02	1.3	0.07	0.036	<0.004	0.0028	835.7°C	835.7°C

* FACT software

** Thermo-Calc software

Fig. 1. Thermomechanical schedule for the torsion tests. In the first deformation, the specimens were strained to 2.0 at a strain rate of 0.4 s^{-1} . In the second deformation, the specimens were strained to 0.5–5.0 at strain rates of 0.04 s^{-1} and 0.4 s^{-1} .

carburization during heating. The specimens were heated at 1°C/s to 1200°C , held for 1 min, and then strained to $\varepsilon=2.0$ at a strain rate of $\dot{\varepsilon}=0.4 \text{ s}^{-1}$ to condition the austenite prior to deformation; after 20 min at 1200°C , the specimens were cooled at 1°C/s to the various test temperatures. Prior to the second deformation, the torsion specimens were held for 1 min at the test temperature; they were then strained to $\varepsilon=2.0$ at strain rates of $\dot{\varepsilon}=0.04 \text{ s}^{-1}$ and $\dot{\varepsilon}=0.4 \text{ s}^{-1}$ in the McGill servo-hydraulic computer-controlled MTS torsion machine. In a separate set of experiments, samples were strained to $\varepsilon=0.5, 1.0, 1.5, 2.5, 3.0$ and 5.0 to follow the progress of the strain-induced transformation in more detail. The largest strain of $\varepsilon=5.0$ was only attained when dynamic recrystallization was taking place. Otherwise the maximum strain was limited by fracture of the specimen. The maximum strains in the absence of DRX in the present series were $\varepsilon=2.9$ at 856°C , $\varepsilon=4.4$ at 886°C and $\varepsilon=4.6$ at 896°C . This point will be taken up in more detail later.

The temperature measurements were carried out with an Omega Super XL type K thermocouple that was placed in contact with the mid-length of the torsion specimen. About 1.5 s after each torsion test, the specimens were quenched in water to arrest the progress of static and post-dynamic recrystallization; this also prevented the static decomposition of austenite into cementite and ferrite during cooling.

2.2. Metallography

Cross-sections perpendicular to the longitudinal axis were cut from the deformed specimens. These were hot mounted in a thermosetting resin for optical microscopy and polished with SiC papers to a 1200 grit surface finish. Both 3 and $1 \mu\text{m}$ diamond pastes were employed in the final polishing, after which the specimens were etched in a 2% nital solution.

Table 2. Key to ferrite phase percentages.

Ferrite (α) %	$0 < \alpha < 20\%$	$21 \leq \alpha < 40\%$	$41 \leq \alpha < 60\%$	$61 \leq \alpha < 80\%$	$81 \leq \alpha \leq 100\%$
Color

The equivalent stresses and strains were calculating using the expressions¹²⁾:

$$\sigma = \frac{3.3\sqrt{3}T}{2\pi r^3} \dots\dots\dots(2)$$

and

$$\varepsilon = \frac{r\theta}{\sqrt{3}L} \dots\dots\dots(3)$$

where σ is the equivalent stress, T the applied torque, r the radius of the specimen, ε the equivalent strain, θ the angle of twist in radians, and L the length of the specimen.

As is evident from Eq. 3, the strain is maximum at the surface of the specimen, while it is zero along the axis. The metallographic observations were therefore conducted mostly near the outer radius. Nevertheless, analyses were also carried out along the radius of the specimens to determine the effect of strain at constant time of deformation.

2.3. Quantification of the Ferrite Phase and Grain Size Measurements

The changes in the percentages of induced ferrite with strain rate, strain, time and temperature were followed by means of metallographic observations. An optical microscope equipped with an image analyzer was used to determine the percentage of ferrite by measuring the ferrite areas and subtracting them from the total area. The dependence of ferrite volume fraction on radius (strain) was measured using a series of optical micrographs extending from the axis of the specimen to the surface; for this purpose, magnifications of $\times 500$ and $\times 1\,000$ were employed. The ferrite percentages were then classified into five groups at intervals of 20%, as shown in **Table 2**. The image analyzer of the optical microscope was used to measure the grain size of the deformation-induced ferrite by standard linear intercept methods.

3. Results

3.1. Microstructure

Some typical metallographic results are presented in **Figs. 2** to **5**. The deformed specimens contained two phases: ferrite and martensite (the latter produced by quenching the austenite after deformation). The displacive transformation of austenite into martensite formed thin plates that had the appearance of “needles”, while the strain-induced ferrite grains were of approximately equiaxed shapes. These morphological differences were employed to distinguish between the two phases. As the strain increases with radius, the deformation-induced ferrite appeared initially at the outer layers of the specimen, generally at strains of $\varepsilon \geq 0.5$, *i.e.*, the critical strain (Fig. 2). On further increase in strain, the amounts of ferrite increased and the ferrite/martensite boundary was gradually displaced towards the inner layers (Fig. 3). Some of the optical micro-

Fig. 2. Optical micrographs showing formation of the ferrite phase by DT in specimens deformed at 856°C ($A_{c3}+20^\circ\text{C}$): (a) $\varepsilon^\circ=0.04\text{ s}^{-1}$, $\varepsilon=0.5$, ferrite and martensite needles (arrows), (b) $\varepsilon^\circ=0.04\text{ s}^{-1}$, $\varepsilon=1.0$, ferrite, (c) $\varepsilon^\circ=0.4\text{ s}^{-1}$, $\varepsilon=0.5$, ferrite and martensite needles (arrows), (d) $\varepsilon^\circ=0.4\text{ s}^{-1}$, $\varepsilon=1.0$, ferrite. Here, the critical strain for ferrite formation was 0.5 and the volume fraction of the ferrite phase increased with strain. At a strain of 1.0, b) and d), all the austenite had transformed to ferrite at both the lower and higher strain rates.

Fig. 3. Cross-sections of specimens deformed at 856°C ($A_{c3}+20^\circ\text{C}$) at a strain rate of 0.4 s^{-1} : a) at a strain of 1.0, the predominant phase is the martensite (needles) formed during quenching; b) at a strain of 1.5, there is a considerable increase in the volume fraction of the ferrite phase.

graphs pertaining to a variety of experiments are presented in Fig. 4.

The occurrence of the reverse transformation during

Fig. 4. Optical micrographs illustrating the ferrite formed by DT under the following conditions: (a) $\dot{\epsilon}^0=0.04 \text{ s}^{-1}$, $\epsilon=2.0$, $T=816^\circ\text{C}$ ($<A_{e3}$); (b) $\dot{\epsilon}^0=0.04 \text{ s}^{-1}$, $\epsilon=2.0$, $T=826^\circ\text{C}$ ($<A_{e3}$); (c) $\dot{\epsilon}^0=0.4 \text{ s}^{-1}$, $\epsilon=2.0$, $T=846^\circ\text{C}$ ($>A_{e3}$); (d) $\dot{\epsilon}^0=0.04 \text{ s}^{-1}$, $\epsilon=1.5$, $T=856^\circ\text{C}$ ($>A_{e3}$); (e) $\dot{\epsilon}^0=0.04 \text{ s}^{-1}$, $\epsilon=2.0$, $T=876^\circ\text{C}$ ($>A_{e3}$); (f) $\dot{\epsilon}^0=0.4 \text{ s}^{-1}$, $\epsilon=5.0$, $T=896^\circ\text{C}$ ($>A_{e3}$).

Fig. 5. Micrographs of specimens deformed at 856°C ($A_{e3}+20^\circ\text{C}$) at 0.4 s^{-1} and held isothermally after deformation for (a) 0 s, (b) 200 s, (c) 400 s. The reverse transformation began after 200 s of holding, as revealed by the presence of the martensite needles (arrows).

holding is illustrated in Fig. 5. These specimens were deformed at the higher strain rate $\dot{\epsilon}^0=0.4 \text{ s}^{-1}$ at 856°C ; after deformation, the specimens were held isothermally for 50, 100, 200 and 400 s.

Fig. 6. Stress-strain curves of specimens deformed at different temperatures at 0.04 s^{-1} to a strain of 2.0. After the peak stress, the flow stress decreases slightly as a result of DT.

Fig. 7. Stress-strain curves of specimens deformed at different temperatures at 0.4 s^{-1} to a strain of 2.0. After the peak stress, the flow stress decreases slightly solely by DT at 816 and 836°C . At 846 , 856 , 866 , 876 and 896°C , the decrease in flow stress after the peak results from the combined effect of DT and DRX.

3.2. Stress-Strain Curves

Some typical stress-strain curves determined during the torsion tests at the lower strain rate ($\dot{\epsilon}^0=0.04 \text{ s}^{-1}$) are depicted in Fig. 6. Representative curves for the higher strain rate ($\dot{\epsilon}^0=0.4 \text{ s}^{-1}$) are illustrated in Fig. 7. The higher flow stress values coupled with the absence of DRX in all the lower strain rate experiments and in the higher strain rate experiments at 816 and 836°C (*i.e.* at or below the A_{e3}) limited the maximum attainable strain to 2.0. The precipitation of niobium carbides on dislocations may have been responsible for suppressing the occurrence of dynamic recrystallization^{13,14} in the lower strain rate experiments as well as in the higher strain rate experiments at 816 and 836°C . Nevertheless, some specimens were deformed to strains above 2.0 at temperatures above 836°C (A_{e3}) even in the absence of DRX. This permitted observation of the flow behavior for longer periods of time (Fig. 8).

During the deformation tests, two complementary softening mechanisms were observed to operate. The first is dynamic transformation (DT). As the strain-induced ferrite is softer than the parent austenite, its formation reduces the rate of work hardening normally displayed by austenite, leading to a decrease in the flow stress and the eventual appearance of a peak (even in the absence of DRX). Softening by DT began once the critical strain was exceeded (about 0.5). However, as has already been pointed out, appreciable volume fractions of the ferrite phase only appeared beyond

Fig. 8. Stress-strain curves of specimens deformed at different temperatures to fracture or to a strain of 5.0. In the lower strain rate tests, the flow stress decreases after the peak solely as a result of DT. In the higher strain rate tests, the flow stress decreases as a result of the combined effect of DT and DRX. The occurrence of DRX permitted the attainment of larger strains.

strains of about 1.0 (see Fig. 3). As a result, flow softening by DT was only detected in the stress-strain curves beyond strains of 1.2–1.7 in the lower strain rate and 1.0–1.6 in the higher strain rate experiments.

The second softening mechanism, DRX, was observed in the higher strain rate experiments at temperatures above 836°C (A_{e3}). DRX was concentrated in the ‘interior’ layers of the specimens, where the strains and strain rates were lower than in the outer layers, so that the critical strain for DT had not yet been reached. The combined effect of DT and DRX on the flow curves at 856 and 896°C can be seen in Figs. 7 and 8.

3.3. Procedure for Evaluating the Ferrite Volume Fractions

The method used to quantify the volume fraction of the ferrite phase is illustrated in **Fig. 9**. This particular specimen was deformed at 856°C at a strain rate of $\dot{\epsilon}=0.4 \text{ s}^{-1}$ to a strain of $\epsilon=2.0$. The amount of strain-induced ferrite increased with radius (*i.e.* strain) and, at the surface, the only phase observed was ferrite as the critical strain of 0.5

Fig. 9. Method used to quantify the amount of the ferrite phase: a) martensite grains formed during the quenching of austenite; some ferrite (arrows) is seen to have precipitated at the grain boundary; b) grains of ferrite (arrows) formed by DT in the martensite matrix; c) martensite grains (arrows) in a ferrite matrix; d) ferrite grains. The specimen was deformed at 856°C at a strain rate of 0.4 s^{-1} to a strain of 2.0. $0 < \text{ferrite} < 20\%$ ■; $21 \leq \text{ferrite} < 40\%$ ■; $41 \leq \text{ferrite} < 60\%$ ■; $61 \leq \text{ferrite} < 80\%$ ■; $81 \leq \text{ferrite} \leq 100\%$ □.

had been exceeded. The core of the specimen was composed of the martensite produced by quenching the austenite present at the test temperature. Close to the center, some ferrite precipitated at the prior austenite grain boundaries (now martensite) during the quenching process (Fig. 9(a)).

Fig. 10. Percentage of ferrite as a function of strain. The specimens were deformed at 856°C ($A_{e3}+20^\circ\text{C}$) to increasing strains at (a) 0.04 s^{-1} , (b) 0.4 s^{-1} . $0 < \text{ferrite} < 20\%$ ■; $21 \leq \text{ferrite} < 40\%$ ■■; $41 \leq \text{ferrite} < 60\%$ ■■■; $61 \leq \text{ferrite} < 80\%$ ■■■■; $81 \leq \text{ferrite} \leq 100\%$ □.

Fig. 11. Percentage of ferrite as a function of temperature. The specimens were deformed to a strain of 2.0 at (a) 0.04 s^{-1} , (b) 0.4 s^{-1} . $0 < \text{ferrite} < 20\%$ ■; $21 \leq \text{ferrite} < 40\%$ ■■; $41 \leq \text{ferrite} < 60\%$ ■■■; $61 \leq \text{ferrite} < 80\%$ ■■■■; $81 \leq \text{ferrite} \leq 100\%$ □.

The dependences of the ferrite volume fractions on strain at 856°C are compared in Fig. 10 for the experiments carried out at the lower and higher strain rates. It is evident that the volume fraction increases with the strain, both along the radius, as well as with the amount of twist. An increase in strain rate from 0.04 to 0.4 s^{-1} also led to detectable increases in the amount of the ferrite phase along the radius.

The influence of deformation temperature on the ferrite volume fraction is depicted in Fig. 11. Here the results obtained at the lower ($\dot{\epsilon}=0.04\text{ s}^{-1}$) and higher ($\dot{\epsilon}=0.4\text{ s}^{-1}$) strain rates are compared for specimens strained to 2.0 at the outer radius. As mentioned above, an increase in strain rate produced a small increase in the amount of strain-induced ferrite produced. The presence of strain-induced ferrite was observed right up to 896°C ($A_{e3}+60^\circ\text{C}$), the maximum temperature employed in these tests (Fig. 12).

The ferrite volume fraction is shown as a function of holding time in Fig. 13 for specimens deformed at the higher strain rate ($\dot{\epsilon}=0.4\text{ s}^{-1}$) and held isothermally after deformation for 50, 100, 200 and 400 s.

It was observed that the reverse transformation took place in two stages: 1) during the initial 200 s of holding, the phase proportions remained fairly stable, 2) after about 200 s of holding austenite began to appear in sites previously occupied by ferrite. This process attained saturation

Fig. 12. Percentage of ferrite in specimens deformed at 896°C ($A_{e3}+60^\circ\text{C}$) to strains of 2.0 and 5.0 at 0.4 s^{-1} . $0 < \text{ferrite} < 20\%$ ■; $21 \leq \text{ferrite} < 40\%$ ■■; $41 \leq \text{ferrite} < 60\%$ ■■■; $61 \leq \text{ferrite} < 80\%$ ■■■■; $81 \leq \text{ferrite} \leq 100\%$ □.

Fig. 13. Percentage of ferrite as a function of isothermal holding time after deformation at 856°C ($A_{e3}+20^\circ\text{C}$) at a strain rate of 0.4 s^{-1} to a strain of 2.0. $0 < \text{ferrite} < 20\%$ ■; $21 \leq \text{ferrite} < 40\%$ ■■; $41 \leq \text{ferrite} < 60\%$ ■■■; $61 \leq \text{ferrite} < 80\%$ ■■■■; $81 \leq \text{ferrite} \leq 100\%$ □.

at 400 s (Figs. 5 and 13).

These results are consistent with the Monte Carlo predictions of Mingming *et al.* (2004), in which an initial steady state is followed by the reverse transformation. The latter is accelerated by the stored energy contained in the deformed ferrite. The reverse transformation was slower in the present Nb-modified material than in the plain carbon steels of the earlier studies.^{2,3,6} This can be attributed in part to the precipitation of niobium, as well as to the solute drag effect of the Nb in solution. Both these effects can be expected to retard the motion of interphase boundaries.

4. Discussion

4.1. Dynamic Transformation

The dependences of the percentages of ferrite on time and strain are shown in **Figs. 14(a)** and **14(b)** respectively. As the strain in a specimen deformed in torsion depends on the radius, at a distance from the axis of 1.57 mm (*i.e.* at 50% of the surface radius), the equivalent strain and strain rate are 1.0 and 0.2 s^{-1} , respectively, in a specimen strained to 2.0 at 0.4 s^{-1} . In this way, it is possible to evaluate the changes in microstructure at different strains and strains rates by analyzing the microstructure at selected radii. In the present study, a radius of 2.42 mm was selected because it satisfied the following conditions: 1) the critical strain for DT was exceeded at this location; and 2) the combination of strain and strain rate at this radius was suitable for determining the kinetics of the transformation.

These studies revealed that, at strains above 0.38, slightly higher percentages of ferrite were present in the specimens

deformed at the higher strain rate. This effect is achieved through the stored energy associated with the dislocations introduced, which increases with the strain rate (at a given strain). At strains above 1.53, the differences in ferrite content decreased with increasing strain, as the transformation approached saturation. The most important observation is that the applied strain has a far higher influence on the transformation than the time.

The measured grain sizes are presented in **Fig. 15**. Here it can be seen that the average grain size increases with temperature and decreasing strain rate over the range 2 to $3.5 \mu\text{m}$.

These results are in general agreement with those of previous workers^{1,2} who found that the transformation of deformed austenite above A_{e3} produced ferrite microstructures in the same size range. In their work, higher strain rates were employed, but the steels did not contain Nb. These two experimental parameters can be seen here to have opposite effects on grain size.

4.2. Industrial Implications

The present results (as well as those of earlier workers) suggest that the strain-induced formation of ferrite takes place quite frequently in the finishing stages of hot rolling. Because the parameters affecting this mechanism are not well known or understood, it has not yet been possible to control it in an industrially useful way. A good candidate for its implementation is in planetary hot rolling, where large reductions are applied in a single operation, which could be carried out just above the conventional A_{e3} . By contrast, this paradigm is not well adapted to plate rolling, because of the large interpass (delay) times involved. Similar remarks apply to Steckel mill rolling. However, it is applicable in principle to strip rolling, where the interpass times are short. Here, the important issues concern the higher rolling loads that could be involved if slightly lower finishing temperatures are to be employed. A further issue concerns mill modifications that would reduce the time before runout table cooling is applied. This would reduce the possibility of reverse transformation, particularly in the case of plain carbon steels. In microalloyed steels, as shown here, the reverse transformation is delayed considerably. Finally, as dynamic transformation leads to a decrease in flow stress, more detailed knowledge of the effects of this transformation should lead to improvements in the ability to

Fig. 14. Dependence of the ferrite percentage on: (a) time, (b) strain. The specimens were deformed at 856°C at both the lower (0.04 s^{-1}) and higher (0.4 s^{-1}) strain rates. The volume fractions of the ferrite phase were measured at a radius of 2.42 mm. At this location, the equivalent strain rates were 0.031 s^{-1} and 0.31 s^{-1} , respectively.

Fig. 15. Mean grain size and standard deviation as a function of strain rate and temperature. The specimens were deformed at both the lower (0.04 s^{-1}) and higher (0.4 s^{-1}) strain rates to a strain of 2.0.

control the rolling load in strip mills.

5. Conclusions

The effect of niobium on the dynamic transformation of austenite to ferrite as well as on the reverse transformation was investigated at a series of temperatures above the A_{e3} . The results were analyzed and led to the following conclusions.

(1) In the present material, the critical strain for dynamic transformation above the A_{e3} is about 0.5. It is clear that the applied strain has a far greater influence on the transformation than the time.

(2) The presence of niobium slows the reverse transformation by comparison with the behavior of plain carbon steels. This probably arises as a result of dislocation pinning by niobium carbonitride precipitation and the solute drag attributable to the presence of Nb in solution. The reverse transformation in the present steel was characterized by two stages. In stage I, which extended over the initial 200 s of isothermal holding, the deformation-induced ferrite remained fairly stable. In stage II, observed after 200 s of holding, the reverse transformation took place and attained saturation after about 400 s.

(3) The retardation of reverse transformation attributed to Nb addition indicates that the latter could play a useful role in preventing reverse transformation during steel processing.

(4) Although equilibrium phase diagrams are essential for understanding the effect of chemical composition on the A_{e3} and A_{e1} transformation temperatures, they do not indicate the 'effective' transformation temperatures that apply to hot rolling operations.

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