Dynamic Transformation of a Low Carbon Steel at Temperatures above the 
Ae₃

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The dynamic transformation of austenite was studied in a 0.06%C low carbon steel by deforming to strains of 0.25–5.0 at strain rates of 0.04 and 0.4 s⁻¹ over the temperature range 877–917 °C. All these temperatures are at or above the conventional Ae₃. Two critical strains were detected, the first (about ε=0.2) was for the formation of strain-induced ferrite; the second (about ε=1.5) was the critical strain for the reverse transformation, which was gradual and only observed in specimens deformed at the lower strain rate. After deformation, the strain-induced ferrite was stable for about 60 s of isothermal holding. However, after this time, the reverse transformation began to take place, approaching saturation in about 240 s. The influence of strain, strain rate and temperature on the dynamic transformation is described. The results indicate that reverse transformation is unlikely to take place in the finishing stands of strip mills due to the high strain rates and short interpass times involved. It is also suggested that conventional phase diagrams do not apply to austenite undergoing deformation in rolling mills.

KEY WORDS: deformation-induced ferrite; dynamic transformation; reverse transformation; Ae₃; hot rolling.

1. Introduction

Due to its industrial importance, the effect of austenite deformation on the formation of ferrite at temperatures below the Ae₃ has been studied by many researchers. Nevertheless, it is only since the 1980’s that researchers have turned their attention to the formation of deformation-induced ferrite (labeled DT in this work) above the conventional Ae₃ temperature.1–10 In the present work, the dynamic transformation of austenite was investigated in a 0.06%C low carbon steel by means of torsion testing. Attention was also paid to the reverse transformation, both during and after deformation. The results obtained are described in detail and then compared with those presented previously9 on the dynamic transformation of a niobium-containing low carbon steel.

The torsion test was employed because it permits much larger strains (up to 5.0) to be applied than the compression method used in some of the investigations.1–3 Such tests also provide more data for analysis, as the dependence of the local strain on radius leads to the availability of multiple observations on a single specimen. The formation of strain-induced ferrite and its reverse transformation during deformation (labeled DRT in this work) as well as after deformation (i.e. static reverse transformation or SRT) were also followed by means of optical microscopy. The effects of strain, strain rate, temperature, niobium content and holding time after deformation on the microstructure are then discussed.

2. Experimental Procedure

Cylindrical torsion specimens 3.15 mm in radius and 22.4 mm in length were machined from a steel plate with the cylinder axes being parallel to the rolling direction. The chemical composition and Ae₃ temperature of the low carbon steel investigated in this work and the previously tested niobium-containing low carbon steel are given in Table 1.

The thermomechanical schedule followed in the torsion tests was similar to the one used in the previous investigation carried out on a niobium-containing low carbon steel.7 The first deformation was applied at 1200°C to condition the austenite and the second at temperatures at and above the Ae₃ so as to explore the phenomenon of austenite transformation above the orthoequilibrium Ae₃. In the second deformation, the specimens were strained in torsion to ε=0.25–5.0 at strain rates of 0.04 and 0.4 s⁻¹ and 10 degree intervals in the range 877–917°C were employed. At 897°C, extra specimens were strained to 5.0 at 0.4 s⁻¹ and held for 15, 30, 60, 120 and 240 s before quenching so as to follow the reverse transformation of ferrite to austenite. After torsion testing, all the samples were water quenched in about 1.5 s; this was done to prevent the static decomposition of austenite during cooling into cementite and ferrite.

Table 1. Chemical compositions (mass%) and equilibrium temperatures Ae₃ of steels investigated.

| C | Si | Mn | Al | Nb | P  | S  | Ae₃ (°C) |
|---|----|----|----|----|----|----|----------|
| 0.06 | 0.01 | 0.3 | 0.042 | — | 0.013 | 0.008 | 877 |
| 0.09 | 0.02 | 1.3 | 0.07 | 0.036 | <0.004 | 0.0028 | 835.7 |
experiments were conducted in an atmosphere of argon plus 2% hydrogen to minimize decarburization.

Cross-sections perpendicular to the longitudinal axis of the deformed specimens were prepared for metallographic observations. These were hot mounted and polished with silicon carbide paper and diamond pastes of 3 and 1 microns in the final stages. After polishing, the specimens were etched with a 2% nital solution.

In specimens deformed by torsion, the strain and strain rate are zero along the axis and they increase linearly from the center to the surface, where they are at their maxima. The amount of deformation-induced ferrite is therefore expected to be at a maximum at the outer surface and to vary along the radius. For this reason, microscopic observations were carried out along the radius to observe the effects of different strains and strain rates on the microstructure. The dependence of strain and strain rate on radius is described in detail elsewhere.7,12)

3. Results
3.1. Microstructure
Some typical optical micrographs are presented in Figs. 1 to 4. The deformed and quenched specimens contain fairly equiaxed ferrite grains (formed by dynamic transformation) as well as the needle-shaped martensite plates formed by quenching. The effect of strain on the transformation is illustrated in Fig. 1 for a deformation temperature of 887°C (Ae3+10°C) and in Fig. 2 for a temperature of 907°C (Ae3+30°C). From Fig. 1(a), it can be seen that the critical strain for ferrite formation is about 0.2 (i.e. it is located at a radius of 2.52 mm in a specimen of radius of 3.15 mm strained to 0.25 at the surface). The critical strain takes a similar value at 907°C, see Figs. 2(a) and 2(c). As the strain is increased to 1.5 (Fig. 2(b)), 2.5 (Fig. 1(b)) and 2.5 (Fig. 2(d)), the proportion of ferrite increases progressively and that of martensite is diminished.

The nature of the strain-induced ferrite can be seen to better effect in the higher magnification micrographs of Fig. 3(a) (887°C) and Fig. 3(c) (917°C–Ae3+40°C). It should be noted that the ferrite grains acquire a nearly equiaxed shape during dynamic transformation at both temperatures. The
characteristics of the martensite produced by quenching the
deformed austenite are illustrated in Fig. 3(b) (887°C) and
Fig. 3(d) (917°C). Here it is evident that the lower strain and
strain rates at or close to the center resulted in little or no
DT, which in turn resulted in a structure composed mostly
of martensite plates.

The effect of holding time after deformation on the
microstructure is illustrated in Fig. 4. These specimens were
strained to 5.0 at 4.0 s\(^{-1}\) at 897°C (\(\text{Ae}_3+20°C\)). After deforma-
tion, the specimens were held at temperature for 15, 30,
60, 120 and 240 s. It can be seen that the deformation-
induced ferrite was stable during the initial 60 s of iso-
thermal holding. When the holding time was increased
beyond 60 s, the reverse transformation began to take place,
approaching saturation in about 240 s.

3.2. Stress-strain Curves
Some typical torsion stress-strain curves are presented in
Fig. 5. During deformation, two “hardening” mechanisms
were observed to operate. The first is conventional work
hardening while the second is reverse transformation (aus-
tenite is harder than ferrite). The latter only took place at the
lower strain rate (\(\varepsilon=0.04\) s\(^{-1}\)), as it involved an incubation
time of about 40 s. Dynamic softening was produced by two
competing mechanisms: dynamic recrystallization and
dynamic transformation.

The stress-strain curves for the lower strain rate (\(\varepsilon=0.04\)
\(\)s\(^{-1}\)) experiments are presented in Fig. 5(a). Work hardening
is the only mechanism acting at strains of 0.2. Beyond the
critical strain for dynamic transformation (see Fig. 1 and
Fig. 2), the latter only took place at 897°C (over a strain interval of 0.3) and 5 s at 907°C (over a strain interval of about 0.2). At the higher strain rate
(0.4 s\(^{-1}\)), Fig. 5(b), softening took place more rapidly (2.5 s
at 897°C and 1.5 s at 907°C) due to the higher dislocation
density. This can be attributed to the higher driving force
for strain-induced transformation in the more rapidly deforming
austenite. The slight increase in flow stress observed at
strains greater than about 1.5 indicated that reverse transfor-
mation was taking place in the specimens undergoing strain-
ing. Here the relatively soft ferrite was being replaced by
harder, dislocation-free austenite.

3.2.1. Competition Between Dynamic Transformation and
Dynamic Recrystallization
It is relevant here that the critical strains for the initiation
of these two mechanisms depend somewhat differently on
temperature and strain rate. As can be seen from Figs. 1(a), 2(a) and 2(c), the critical strain for ferrite formation is relatively insensitive to these two experimental parameters. All that is required is for the austenite to contain enough dislocations for its free energy to be raised beyond that of undeformed ferrite at the relevant temperature. By contrast, the critical strain for the nucleation of dynamic recrystallization requires the dislocation substructure to attain a specific configuration that is much more dependent on temperature and strain rate. As a result, dynamic transformation is nucleated in the outer layers of the specimens (where the strain rate is higher) before dynamic recrystallization, whereas the converse condition is likely to apply in the interior layers.

The progress of dynamic recrystallization cannot be followed directly in the present material because of the phase changes that take place, although its general features are known from numerous experiments. It can therefore only be conjectured that, whereas dynamic transformation cannot take place in a newly recrystallized grain (because of the absence of dislocations), such grains only have to be submitted to modest further strains (e.g. 0.2) before they can also participate in dynamic transformation.

Figure 5(c) is included for comparison purposes. Here are displayed the stress-strain curves of specimens of a niobium-containing low carbon steel deformed at different temperatures to fracture or to a surface 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 dynamic transformation and dynamic recrystallization.

3.3. Ferrite Volume Fraction

The method for the quantification of the ferrite volume fraction used here is presented in Fig. 6. Further details are...
provided in our earlier publication.\(^7\) The evolution of the ferrite volume fraction with strain determined in this way in the lower strain rate experiments is presented in Fig. 7(a). The amount increased rapidly with strain over the range 0.2–1.5. At strains beyond 1.5, there is some evidence supporting the view that reverse transformation is taking place, just perceptibly reducing the amount of deformation-induced ferrite.

The evolution of the ferrite volume fraction in the higher strain rate experiments is illustrated in Fig. 7(b). The strain rate increase produced thicker layers of deformation-induced ferrite, compare Figs. 7(a) and 7(b). The increase in ferrite volume fraction with strain rate can be attributed to: a) the higher driving force (i.e. stored energy) for the transformation produced at the higher strain rate; and b) the decrease in the time available for reverse transformation. Under these conditions, the amount of deformation-induced ferrite increased continuously with strain.

Figure 7(c) is included for comparison purposes. It illustrates the dependence of the volume fraction of deformation-induced ferrite on strain in the previous niobium-containing low carbon steel.\(^7\) Here it can be seen that the addition of 0.036% Nb retards the formation of the ferrite phase to a considerable degree (see Figs. 7(a) and 7(c) and Figs. 7(b) and 7(d)), although it does not modify the main features of this phenomenon. Moreover, the addition of niobium completely suppressed the reverse transformation in the lower strain rate experiments (see Figs. 7(a) and 7(c)).

The effect of temperature on dynamic transformation is presented in Fig. 8. Here it is evident that the amount of ferrite formed as well as the driving force for the transformation decrease as the temperature is increased. This trend was observed to hold up to the highest temperature of 917°C. Nevertheless, the present results suggest that it should be possible to produce deformation-induced ferrite at temperatures considerably higher than 917°C. For example, Yada and coworkers\(^2\) reported that they were able to form deformation-induced ferrite at temperatures as high as 1300 K (1026°C) (i.e. \(\text{Ae}_3+193\) K) in their low carbon steel containing 0.14%C–0.33%Si–1.06%Mn.

The effect of holding time on ferrite volume fraction is illustrated in Fig. 9(a) for the present niobium-free steel. The behavior of the niobium-containing steel is presented in Fig. 9(b) for comparison. In the niobium-free low carbon steel, the amount of deformation-induced ferrite remained fairly constant over the initial 60 s. Beyond this time,
reverse transformation occurred gradually over the interval 60–120 s, and approached saturation beyond holding times above 120 s. These results are in good agreement with the Monte Carlo simulations of Tong et al.\(^4\) in which an initial incubation period precedes the reverse transformation to austenite. In the microalloyed material, a similar two-stage reverse transformation process is observed; however, the solute drag effect of niobium retarded the reverse transformation considerably, Fig. 9(b).

4. Discussion

4.1. Forward and Reverse Dynamic Transformations

The increase in stored energy attributable to concurrent deformation (i.e. in the driving force for dynamic transformation above the \(\text{Ae}_3\)) and decrease in deformation time (i.e. in the time available for diffusion during deformation) prevented the reverse transformation from taking place when the strain rate was raised from 0.04 to 0.4 s\(^{-1}\). This suggests that there is a critical strain rate above which the dynamic reverse transformation cannot occur. In the present material, this critical strain rate was about 0.1 s\(^{-1}\). The kinetics of the forward transformation are compared in Fig. 10 for the low and high strain rate experiments. As mentioned earlier, the strain and strain rate in torsion specimens increase linearly with the radius. Thus the effects of different strains and strain rates can be assessed by examining the microstructure at different radii. In this diagram, a radius of 2.2 mm (at which the strain rate is 0.028 s\(^{-1}\) at the low strain rate and 0.28 s\(^{-1}\) at the high strain rate) was chosen to reveal the influence of strain and strain rate on the forward and reverse transformation. After a strain of 1.0, the transformation approached saturation at the higher strain rate. This suggests that an accumulated strain of 1.0 should be sufficient to produce extensive dynamic transformation during hot rolling. By contrast, a saturation value of only about 50% volume fraction was attained at the lower strain rate. This is because the reverse transformation began after a strain of about 1.0 (about 40 s) in these experiments. Thus it appears that the continued straining of deformation-induced ferrite raises its free energy and thus promotes the reversion to austenite.

From Fig. 10, it is evident that the forward transformation is driven by the strain and is not particularly time dependent. In this way, the kinetics rather resemble those of dynamic recrystallization, which are also highly sensitive to strain rate and relatively insensitive to time.

Nevertheless, the application of equal strains at two different radii (e.g. 3 mm and 1.5 mm), see Fig. 7, does not lead to the production of equal amounts of strain-induced ferrite. The strain and strain rate increase linearly with radius but not the amount of ferrite. This is probably because dynamic recovery and DRX in the interior layers are annihilating dislocations and reducing the driving force for the dynamic transformation. However, the detailed effect of these softening mechanisms on dynamic transformation is not understood at the moment and will require further investigation.

4.2. Industrial Implications

The present work demonstrates that there is a critical strain rate for the suppression of the reverse transformation, which is well below that applicable to industrial hot rolling, where much higher strain rates are commonly used. Because ferrite is softer than austenite, dynamic transformation is expected to have an effect on the roll separation force. The addition of niobium to low carbon steels considerably retards both the forward and reverse transformations. As the finishing temperatures in strip mills usually fall in the range 850–925°C, the present observations suggest that new hot rolling paradigms should be developed that take this phenomenon into account. Furthermore, new sets of phase diagrams applicable to dynamic conditions should also be derived that allow for transformation above the conventional \(\text{Ae}_3\) and \(\text{Ae}_1\).
5. Conclusions

(1) Depending on the strain and strain rate, two hardening mechanisms, work hardening and reverse transformation, can operate concurrently during hot rolling above the $\text{Ae}_3$. In a similar manner, there are two concurrent softening mechanisms, dynamic recrystallization and dynamic transformation, that can act to reduce the flow stress.

(2) The critical strain for dynamic transformation above the $\text{Ae}_3$ in low carbon steels is about 0.2. This is appreciably lower than in niobium-containing steels, where the former is 0.5.3,7)

(3) There is a critical strain rate (and time) for the reverse to austenite. In this particular steel, it is about $0.1 \text{s}^{-1}$, which corresponds to an incubation time of about 10 s.

(4) The present results suggests that reverse transformation is unlikely to take place in industrial mills, due to the relatively high strain rates (and short deformation times) involved. In a similar manner, the short interpass times are likely to suppress static reverse transformation.

(5) Because dynamic transformation is probably taking place during deformation in the finishing stands of strip mills, models of hot rolling should be modified to take into account the effect of dynamic transformation on the roll force.

(6) The addition of niobium slows dynamic transformation and both the dynamic and static reverse transformations.

(7) Equilibrium phase diagrams are not appropriate for predicting the occurrence of dynamic transformation above the $\text{Ae}_3$. Diagrams applicable to austenite that is undergoing deformation are required that will eventually permit the development of more accurate models of hot rolling.

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