Introduction

High strength steel with high toughness is the main objective of thermomechanical processing. There are different approaches to achieve these requirements, although refinement of the ferrite grain size is the simplest method to increase both strength and toughness. Historically, ferrite refinement has been achieved in a number of ways by thermomechanical processing. However, it has been shown that there is a limiting ferrite grain size of approximately 5 \( \mu \text{m} \) regardless of the level of retained strain induced into the austenite using conventional thermomechanical processes such as controlled rolling.\(^1\) This is at least partly due to ferrite coarsening during transformation.\(^2\)

More recently, researchers have developed new thermomechanical processes to produce fine, equiaxed ferrite grains smaller than 2 \( \mu \text{m} \), here called ultrafine ferrite (UFF), in as hot rolled steel strip by different mechanisms. Basically, the potential routes to produce UFF may be classified into three categories.

The first method uses the austenite to ferrite transformation to obtain UFF from a fine prior austenite grain structure, in which the fine austenite grain size is produced by recrystallisation\(^3\) or other processes.\(^4\) Secondly, recrystallisation in the ferrite phase field can be used to produce UFF,\(^4,5\) although the stacking fault energy of ferrite is high, which leads to sluggish dynamic recrystallisation of ferrite. The third method uses strain induced transformation of austenite to ferrite (SIT) in conjunction with rapid cooling. Hodgson and colleagues\(^6\) developed a new thermomechanical process based on SIT, which produces UFF as small as 1 \( \mu \text{m} \) in the surface of hot rolled steel strip. In view of the continuing interest in achieving even more efficient grain refinement by the latest thermomechanical processing, the present work evaluates the effect of strain induced transformation on ultrafine ferrite and bainite grain structure formation through single pass rolling in a plain carbon steel. Furthermore, the microstructural evolution of ferrite transformation is studied through the strip thickness at the critical strain for UFF formation.

Experimental procedure

The composition of steel used in this study is given in Table 1. Hot rolling was performed on a laboratory mill with rolls of 365 mm diameter, a rolling speed of 15 rev min\(^{-1}\) and preheating resistance furnace with a maximum working temperature of 1300°C. The wedge samples were machined to 140 \( \times \) 20 \( \times \) 8 mm with an included angle of about 4° (Fig. 1). All the specimens were austenitised at 1200°C for 15 min, resulting in an average austenite grain size of \( \sim 120 \mu \text{m} \) (Fig. 2). The samples were reheated in stainless steel foil bags to prevent excessive oxidation. The wedge samples were cooled to the \( \text{Ae}_1 - \text{Ar}_1 \) region between Kaowool blankets to homogenise the temperature along the length of the wedge. The specimens were then passed through the rolls at a temperature of 760°C to achieve strip with a final thickness of \( \sim 2 \) mm. The temperature of the specimens was measured continuously during the test by two N-type thermocouples, inserted at opposite positions in the sides of samples (Fig. 1). The surface temperature was measured by a pyrometer throughout the experiment. After deformation, the rolled strips were air cooled to room temperature. To study the microstructural evolution of the ferrite transformation through the thickness, the air cooling was interrupted by water quenching from a given temperature.

<table>
<thead>
<tr>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>Al</th>
<th>( \text{Ae}_3 ), °C*</th>
<th>( T_{\text{air}} ), °C</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.35</td>
<td>0.26</td>
<td>0.82</td>
<td>0.03</td>
<td>795</td>
<td>780</td>
</tr>
</tbody>
</table>

*\( \text{Ae}_3 \) temperature was calculated using the ChemSage program.
Metallographic observations were made on the normal rolling direction plane at the centre and at a depth of between 0.1 and 0.15 mm below the surface of the rolled strips. The ferrite volume fraction was determined by point counting. In this study, only the volume fraction of polygonal ferrite was considered. Electron backscattered diffraction (EBSD) examination was used to distinguish the ferrite grain boundaries with misorientation angles more than 15° using HKL Technology Channel 5.* The mean ferrite grain size was determined using the mean linear intercept method.

Results

Due to the changing thickness along the length of the wedge, there is a temperature variation in the wedge sample after the air-cooling before deformation (Fig. 3). Due to this, the thinner sections (<4 mm) of the wedge sample were deformed in the two phase region (i.e. lower than the Ar3 temperature) rather than in the austenite single phase region. The final thicknesses of the deformed strips were typically in the range 2 to 2.3 mm. The microstructures of all rolled samples were inhomogeneous through the strip thickness. The surface microstructure showed two transition points (Fig. 4).

Transition I: change from a ferrite-pearlite to a bainite microstructure. Transition II: change from a bainitic to an UFF microstructure.

The critical reduction and deformation temperature $T_d$ for each transition are given in Table 2. Transitions I and II began at 40% and 70% reduction, respectively (Fig. 4). However, because the temperature also varied along the length it is not possible to only specify a critical reduction, but a combined critical reduction and deformation temperature.

**TRANSITION I (BAINITIC TRANSFORMATION)**

At the location corresponding to 40% reduction the surface layer studied (0.1–0.15 mm depth) consisted of very fine bainite. The measured surface temperature was $580 \pm 10 ^\circ C$ just after the exit points of the roll gap. A much lower surface temperature would have existed within roll gap, as the surface temperature rapidly increases through conduction from the hotter strip centre (Fig. 5). At these lower temperatures, the available CCT data for a similar steel is consistent with the formation of bainite. However, it is important to note the extreme fineness of the bainite formed here compared with other bainite structures formed after thermomechanical processing.

**TRANSITION II (UFF FORMATION)**

The main microstructural feature after the critical strain for UFF formation was the presence of equiaxed ferrite grains as small as $1.5 \pm 0.25 \mu m$. These penetrated a quarter of the total thickness from each side, with some evidence for carbides on grain boundaries and grain interiors (Fig. 6). The strip core had a more conventional microstructure of ferrite and pearlite.

Microstructural evolution of UFF transformation in the surface layer

At 70% reduction, very fine ferrite grains completely decorated the prior austenite grain boundaries and nucleated into closely spaced parallel arrays or three dimensional rafts within each austenite grain at an early stage of transformation (Fig. 7a). The distribution of the ferrite grain size was very inhomogeneous. There were a few carbides in the vicinity of some ferrite grains, which may have been precipitated during quenching.

With a decrease in the quench temperature, the volume fraction of ferrite increased through ferrite growth and the nucleation of new, intragranular ferrite grains in the remaining austenite (Fig. 7b and c). The intragranularly nucleated grains had grown significantly compared with the ferrite grains nucleated on or near the prior austenite grain boundaries. The distribution of the ferrite grain size was reasonably homogenous with more than 75% ferrite grains less than 2 μm in size (Fig. 8). Also, the volume fraction of carbide increased markedly. These carbide particles were arranged in elongated and narrow bands; their orientations did not appear to have any specific relationship with the

**Table 2 The critical reduction and deformation temperature for each transition**

<table>
<thead>
<tr>
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<th>Transition I</th>
<th>Transition II</th>
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<tbody>
<tr>
<td>$T_d$, °C</td>
<td>710</td>
<td>751</td>
</tr>
<tr>
<td>Reduction (%)</td>
<td>40%</td>
<td>70%</td>
</tr>
</tbody>
</table>

*HKL TECHNOLOGY APS, Denmark, July 2001.
rolling direction (indicated by the arrow in Fig. 7b). However, the density of carbides along the prior austenite grain boundaries was much higher than in the austenite grain interiors. These bands were not evident in the steel prior to deformation.

Ferrite microstructure in the centre
At the sample location corresponding to 70% reduction, the ferrite grain size was much coarser in the centre compared with the surface at any given quench temperature (Fig. 9). The volume fraction of transformed ferrite also decreased from the surface to the centre of the strip (Fig. 7). The density of intragranular rafts was clearly reduced since no intragranular ferrite grains were seen in the centre of the strip at a quench temperature of 675°C (Fig. 7d). Also, the ferrite had not completely decorated the prior austenite grain boundaries in the centre of the strip (Fig. 7d). This variation in volume fraction of ferrite from surface to centre is due to a change in the strain gradient and undercooling through the thickness. With a decrease in the quench temperature, the ferrite grain size coarsened and some intragranular ferrite nucleated. The intragranular ferrite had a more acicular morphology (indicated by the arrow in Fig. 7e). Furthermore, at the lower temperature, pearlite had formed in the centre of the strip instead of fine carbides (Fig. 7f). The volume fraction of acicular ferrite and pearlite increased in the centre compared with the region closer to the surface.

The density per unit boundary length $N/L$ of ferrite nucleated on or near the prior austenite grain boundary was reduced for an increase from 50% to 60% reduction, with an associated increase in the ferrite grain size (i.e. through progress of transformation), after cooling to room temperature (Fig. 10). However, for 70% reduction $N/L$ increased at the start of transformation, followed by a slight decrease with increase in ferrite grain size. Microstructural studies showed that the prior austenite grain boundaries were completely decorated by ferrite grains for the 50% reduction at an early stage of transformation. However, the majority of prior austenite grain boundaries were free of ferrite grains at 70% reduction. This can be explained because the deformation temperature and initial thickness were different for each reduction, resulting in a different amount of undercooling. Moreover, $N/L$ was slightly lower for boundaries that were parallel to the rolling direction. This difference was higher at the start of transformation and decreased during transformation.
Discussion

The single pass rolling and subsequent quenching or air cooling of the wedge sample showed a number of important features:

1. For less than 40% reduction there was no dynamic strain induced transformation in the surface layer and after air cooling the entire thickness transformed to a conventional ferrite and pearlite microstructure;

2. For locations corresponding to between 40 and 70% reduction, the surface layer showed evidence of a highly refined bainitic microstructure that had formed during deformation (i.e. dynamically). The centre microstructure consisted of a conventional ferrite and pearlite microstructure.

3. For higher than 70% reduction there was dynamic strain induced transformation of ferrite in the surface layer and the air cooled microstructure showed ultrafine ferrite grained structures as small as 1.5 μm.

In the following sections, the evolution of the phases is discussed in more detail.

BAINITIC TRANSFORMATION

In this study, a condition is created in the wedge sample whereby the steel passes through the bainitic transformation region during thermomechanical processing. The as quenched microstructure after deformation shows that the very fine bainitic ferrite was dynamically transformed in the surface layer during single pass rolling (i.e. through SIT).
and the remaining austenite transformed during cooling to room temperature.

So far, the effect of deformation on the bainitic transformation has been studied for alloy steels using either an ausforming process or controlled thermomechanical processing. These processes allow the strain to be accumulated before the bainitic transformation. However, in plain carbon steels, the restoration processes (i.e., recrystallisation and recovery) reduce the strain that can be retained in the austenite before bainitic transformation. In addition, the size of bainitic ferrite is controlled by the density of nucleation sites rather than the growth rate. In this study, the deformation temperature required for very fine bainite formation was very low in comparison with conventional thermomechanical processing (Table 2). The low deformation temperature has been reported to significantly increase the volume fraction of intragranular defects (i.e., deformation bands), which are potential additional nucleation sites for the ferritic bainite. It is proposed that the condition for the formation of very fine bainite formation is very similar to that required for UFF formation. This result suggests the potential for other strain induced transformations of other phases to be formed, although significantly more work is required at this stage.

**UFF FORMATION IN SURFACE LAYER**

Previously, it has been reported that a massive type transformation was induced through strain induced transformation of ferrite. As the massive ferrite transformation is known to occur just below the temperature \( T_0 \) (the temperature at which \( G_e = G_s \)) in iron alloys with very low carbon content (as low as 0.02 wt-%), this mechanism is not easily accepted for UFF formation. Recent work has confirmed that UFF formation through SIT is a diffusional transformation (i.e., nucleation and growth), although the kinetics of transformation are extremely rapid compared with conventional ferrite transformation. The following sections consider the nucleation and growth aspects of UFF and compare these with the more typical grain boundary nucleation as seen in the centre of the strip.

**Nucleation**

It is now known that inducing intragranular defects (i.e., deformation bands) during deformation provides the nucleation sites for the subsequent dynamic transformation and that this is the main reason for UFF formation. A large prior austenite grain size has a greater tendency to develop these bands, the inhomogeneity in strain distribution at a microscale increases with increase in austenite grain size. Also, there is a much higher shear strain localisation in the vicinity of the austenite grain boundaries than the grain interiors. It is proposed that this explains why the density of ferrite rafts in the vicinity of the prior austenite grain boundaries is much higher than the grain interiors in the early stage of transformation (Fig. 7a). However, in the current work the size of ferrite formed near the prior austenite boundary, hereafter termed grain boundary ferrite (GB), and intragranular ferrite (IG) grains were similar at an early stage of transformation (Fig. 7a).

The inhomogeneity of strain also causes some areas of the austenite grains to not have the appropriate driving force for nucleation during deformation. These areas gave coarse austenite islands that remained after deformation and either transformed to martensite during quenching or to ferrite and pearlite on air cooling to room temperature.

**Growth**

The growth of UFF could be divided into two periods: dynamic growth (i.e., during straining) and static growth (i.e., after deformation). Umemoto et al. have shown that the \( \Delta G \) maximum (\( \Delta G^* \)) and the critical radius (\( r^* \)) for ferrite nucleation on dislocations is smaller than that for homogeneous nucleation. The critical radius has been estimated to be about 0.08 \( \mu \)m in size. However, even for rapid
quenching immediately after deformation the mean ferrite grain size is nearly 1 μm. Therefore, the ferrite embryo would have grown very rapidly as it reached the critical size at an early stage of transformation during deformation (here called dynamic growth). On the other hand, if the ferrite structural studies suggest that all nucleation sites at the same time, it is possible that some nucleation sites would be consumed through rapid growth of ferrite nuclei that have nucleated earlier. Hence, potentially not all possible nucleation sites of ferrite could be utilised through strain induced transformation. This may limit the final ferrite grain size, although this strongly depends on the growth behaviour of ferrite grains during the post deformation cooling (i.e. static growth).

Figure 11 shows the ferrite grain size as a function of the fraction transformed in the surface layer. The results are compared with a calculation (equation (1)) based on the assumption that all grains nucleated during transformation become a grain in the final microstructure.\\n
$$d = d_a(X_i/X_c)^{\frac{1}{3}}$$\\

where $d_a$ is the mean linear intercept grain size of ferrite after complete transformation; $X_i$ is the volume fraction of ferrite then present (the remainder being pearlite, martensite or and carbide); and $d_i$, $X_i$ represent corresponding quantities at any instant during incomplete transformation.

It can be seen that the measured mean values for the ferrite grain size (Fig. 11) at each instant of transformation are slightly lower than those predicted by equation (1) from the final grain size. This difference is greater at the early stage of transformation (less than volume fraction of 0.5).

Microstructural studies show that the grain growth behaviours of the IG and GB ferrite are significantly different during transformation in the surface layer of the strip (Fig. 7). Hence, it is necessary to differentiate the IG and the GB grains and study their growth behaviour separately. To estimate the IG and the GB ferrite grain sizes at each stage of transformation, a ferrite grain size of 1 μm was considered as the final grain size of the GB ferrite. The microstructural studies suggest that these grains neither coarsened nor grew normally during post deformation cooling (Fig. 7a–c). With a decrease in the quench temperature, the population of ferrite grains with a size less than 1 μm decreases, comprising approximately 43% of the total population of the final microstructure at room temperature (Fig. 12). This amount is assumed to be the population of the GB ferrite grains.

The GB mean ferrite grain size is approximately constant (0.7 ± 0.1 μm) during transformation (Fig. 13). The estimated mean ferrite grain size values for the IG ferrite grain size at each instant of transformation are lower than those predicted by equation (1) from the final grain size for that population (Fig. 13). This difference is slightly greater at the early stage of transformation. It seems that the assumption of site saturation and normal growth are not completely fulfilled in the IG ferrite grains during the overall transformation, although the amount of coarsening is much lower than in other work\textsuperscript{2,16} that studied the evolution of the ferrite transformation after controlled rolling.

Possible explanations for the difference in grain growth behaviour of the IG and the GB ferrite grains could be the carbide precipitate distribution or just the simple full impingement of grains at an early stage of transformation in the GB ferrite. As stated previously, the strain concentration in the vicinity of the austenite grain boundaries leads to a more rapid transformation in that region. The ferrite grains form a layer around the boundary with full impingement in these dimensions. The remaining small islands of austenite could rapidly transform to carbide during cooling and then impede any further coarsening by normal grain growth through pinning. Also it has been suggested previously\textsuperscript{2} that coarsening of fully impinged rafts could be quite different to coarsening of two-dimensionally impinged rafts where the remaining dimension is still moving through transformation. For the IG ferrite there is not complete impingement when the grain size is smaller and it appears that there is coarsening as the large austenite islands transform to ferrite during cooling.

The hardenability and the intragranular defects play an important role in producing UFF through SIT.\textsuperscript{12,13} Although the coarse prior austenite grain size increases the hardenability of steel and the density of intragranular
defects during straining, it significantly raises the inhomogeneity of the strain distribution at the microscale. In this study, the distribution of strain is significantly inhomogeneous in comparison with other studies resulting in the inhomogeneous distribution of ferrite grains through the microstructure. In fact, the coarse prior austenite grain size could have a negative effect on ferrite refinement through this inhomogeneity of strain distribution. Therefore, there could be an optimum prior austenite grain size that depends on the thermomechanical processing as well as steel composition to decrease the strain inhomogeneity and refine the ferrite grain size effectively.

**FERRITE GRAIN GROWTH BEHAVIOUR IN THE CENTRE**

The rate of ferrite coarsening in the strip centre was much higher than the surface of strip, although it reduced with an increase in strain (Figs. 9 and 10). The lack of carbide precipitates in the centre could be one reason for the increase in the rate of coarsening. A recent study, however, showed that a 1 µm ferrite grain size has been produced in low carbon steel (0.0022 wt-%C) where the volume fraction of carbide was very low. It was proposed in that work that the distribution of misorientation of ferrite grains would presumably be changed through SIT, which could prevent the ferrite coarsening behaviour, although there is no evidence to confirm this proposal. On the other hand, it has been reported that the K–S orientation relationship between the austenite and ferrite was altered by applying strain during transformation. It was claimed that the changes in distribution of the misorientation angle of the ferrite makes it difficult for the ferrite grains to coalesce during transformation and this is responsible for the decrease in aspect ratio of ferrite formed under external deformation. Hence, different ideas have been proposed to explain the reason for ferrite coarsening at the early stage of transformation. In the current work there is ferrite formed dynamically and statically at the surface and statically at the strip interior and it is important to understand the different grain growth behaviour through the thickness.

The amount of strain and undercooling are different at the surface in comparison with the centre of the strip. At the critical condition for UFF formation in the surface (70% reduction) the equivalent mean tensile strain is 1.4. However, the distribution of strain is not uniform through the thickness. To estimate the strain value at different depths from the surface, equation (2) has been used, although it was developed for compression testing:

\[
e = \ln(d_c/TH_e)
\]

where \(e\), \(d_c\) and \(TH_e\) are strain, prior austenite grain size (µm) and thickness of deformed austenite grain (µm), respectively. Figure 14 shows the estimated strain and the mean ferrite grain size as a function of depth from the surface after 70% reduction. This shows that strain at the surface is nearly three times that at the centre of the strip, which agrees with earlier finite element modelling of strip rolling under heavy shear. The high strain accumulation in the surface plays an important role in inducing significant intragranular defects for nucleation of ferrite. However, there is some evidence that shows coarsening still occurs at the surface even in small areas of the microstructure (the IG ferrite grains) as discussed previously.

The other difference between the centre and the surface is the amount of undercooling through contact of the rolls and the sample during deformation. The undercooling at the surface and the centre was 120 °C and 60 °C, respectively, at the 70% reduction location. Grain coarsening kinetics depend strongly on the temperature. Therefore, this undercooling difference could effectively change not only the nucleation rate of ferrite, but also the coarsening behaviour of ferrite through the thickness during post-deformation cooling. It has been reported that any delay between deformation and cooling can cause a significant coarsening in ferrite rafts produced through SIT using hot torsion. In fact, post-deformation cooling can play an important role in controlling the rate of coarsening. This means that the SIT is a necessary but not a sufficient factor to completely control ferrite coarsening in overall transformation while undercooling is a complementary factor to reduce grain growth.

**Conclusions**

Wedge samples were used to study the effect of strain induced transformation on the formation of ultrafine ferrite (UFF) grain structures through single pass rolling. Two transition strains for the bainitic transformation and ultrafine ferrite formation were observed in a plain carbon steel (0.35%C). The bainite and UFF were formed by strain induced transformation at reductions of 40% and 70%, respectively in the surface layer of the strip. The bainite microstructure was very fine. The ultrafine ferrite grain size was 1.5 µm and its distribution was reasonably homogeneous with more than 75% of the ferrite grains less than 2 µm.

The size of prior austenite grain boundary ferrite (GB) and intragranular ferrite (IG) grains was similar at an early stage of transformation at the critical strain for UFF formation (i.e. 70% reduction), although the distribution of ferrite nucleation sites was significantly inhomogeneous throughout the surface layer. In the surface, the ferrite coarsening was markedly reduced through SIT combined with rapid cooling in comparison with the centre of strip, where only conventional static transformation occurred. In the surface, ferrite coarsening mostly occurred in the intragranularly nucleated ferrite grains rather than those on, or near, the prior austenite grain boundaries. Significant ferrite coarsening was observed in the ferrite grains nucleated on the prior austenite grain boundaries in the centre of rolled strip.

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References

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