Effect of strain path on dynamic strain-induced transformation in a microalloyed steel

L. Sun a,⇑, K. Muszka a,b, B.P. Wynne a, E.J. Palmiere a

a Department of Materials Science and Engineering, The University of Sheffield, Sir Robert Hadfield Building, Mappin Street, Sheffield S1 3JD, UK
b Faculty of Metals Engineering and Industrial Computer Science, AGH University of Science and Technology, Mickiewicza 30, 30-059 Krakow, Poland

Received 24 May 2013; received in revised form 20 November 2013; accepted 21 November 2013
Available online 8 January 2014

Abstract

Dynamic strain-induced transformation (DSIT) in a low-carbon microalloyed steel was studied by hot cyclic torsion to understand the interactions between DSIT and strain path reversals, and the subsequent microstructure evolution when subjected to continuous cooling. The critical strain for DSIT (\( \varepsilon_{\text{c,DSIT}} \)) can be determined by analysing the dynamic softening of the flow stress–strain curves. When deformed to the same total accumulative strain of 2.0, deformation with a small strain amplitude in each pass and multiple strain path reversals led to the suppression of DSIT compared to the extensive DSIT ferrite produced by deformation with a large strain amplitude and a single strain reversal. The results reveal that the amplitude of monotonic strain, not the total accumulative strain, in relation to \( \varepsilon_{\text{c,DSIT}} \) determines the occurrence of DSIT. The suppression or promotion of DSIT can be attributed mainly to the increment of the austenite grain boundary area associated with deformation, especially the development of serration and bulging, and, to a lesser extent, to the generation of high-angle boundaries by austenite grain subdivision. The evolution of these planar defects, which are believed to be the primary ferrite nucleation sites, is strongly influenced by strain path changes, and lead to significantly different DSIT behaviours. DSIT ferrite also showed very limited coarsening after continuous cooling as the ongoing deformation produces further nucleation sites in the austenite matrix and causes orientation variation of the DSIT ferrite inherited from austenite parent grains. Based on these observations, it is believed that the transformation mechanisms for DSIT are essentially the same as the reconstructive mechanism during static phase transformations.

1. Introduction

The most widely utilized method of achieving grain refinement through thermomechanical controlled processing (TMCP) of steels is to maximize the number density of nucleation sites in unrecrystallized austenite to produce a fine ferrite grain size after phase transformation. From a practical point of view, the ultimate ferrite grain size achievable by this method is approximately 5 \( \mu \text{m} \) [1]. In recent years, considerable international research and development efforts have been undertaken to develop novel processing methods for producing steels containing ultrafine ferrite (UFF, \( \sim 1 \mu \text{m} \) grain size) or “very fine” ferrite (VFF, 2–3 \( \mu \text{m} \) grain size) [2–4]. Heavy deformation of super-cooled thermodynamically metastable austenite, i.e. below the \( A_{\text{e3}} \) but above the \( A_{\text{r3}} \) temperatures, is a prominent example that has received significant attention because it requires only relatively simple TMCP routes [5,6]. It is generally believed that the formation of UFF/VFF is assisted by deformation-induced austenite-to-ferrite phase transformation [7,8]. Several names have been given to this refinement mechanism, including deformation-induced ferrite transformation, strain-induced transformation, strain-induced dynamic transformation.
and dynamic strain-induced transformation (DSIT). Some researchers used the term “dynamic” as the ferrite grains are believed to be formed dynamically during deformation of the austenite [7,9]. Work by Yada et al. [10–12] using in situ X-ray diffraction showed that the (111) \( \gamma \) peak appeared concurrently with the (110) \( \alpha \) diffraction peak when deforming several Fe-Ni-C steels in torsion, at temperatures above the \( A_{c3} \) and near to the \( A_{c1} \). This technique provided direct evidence that the \( \gamma \to \alpha \) phase transformation occurred dynamically during deformation. Therefore, the term DSIT will be used in the present work for this type of phase transformation. Recently, several research groups reported that DSIT could even occur at temperatures above the \( A_{c3} \) [13,14]. This could be attributed to the plastic deformation effectively raising the \( A_{c3} \) temperature for deformed austenite by reducing the energy barrier for nucleation and increasing the driving force for transformation. Nevertheless, the dynamic nature of such transformation was confirmed by the observation of dynamic flow softening and microstructure deformation in ferrite grains [15].

DSIT phenomena were reported as early as the 1980s by various researchers [17–20]. More recently, the potential for utilizing the DSIT mechanism to produce UFF steels was exploited in strip rolling by Hodgson et al. [21,22]. This sparked extensive systematic studies of the DSIT mechanisms and the effects of processing parameters such as deformation mode, strain, strain rate, deformation temperature and post-deformation cooling rate on the formation of UFF through the DSIT process. Two very recent review papers, by Beladi et al. [7] and Dong et al. [23], have summarized the latest understanding of the DSIT process. One common agreement is that there are two critical strains: \( c_{DSIT} \), the threshold strain for the onset of DSIT, which has physical meaning as the starting point of dynamic transformation; and \( c_{UFF} \), which is the minimum strain needed to obtain a UFF microstructure after continuous cooling. This latter critical strain is related not to any physical event but, rather, to processing parameters, specifically cooling rate.

Most of the published work on DSIT has been conducted in the laboratory using large monotonic compression or torsion. The required low deformation temperatures, coupled with large deformation (50–80% reduction), present enormous challenges for the industrialization of DSIT to produce UFF steels. For example, the required rolling force to achieve such monotonic deformation far exceeds the capability of current hot rolling technologies [24]. Therefore, practical industrial approaches inevitably require multi-pass deformations which involve strain path changes. To maximize the DSIT effect on grain refinement through multi-pass deformation, it is necessary to gain a fundamental understanding of the interaction between the strain path changes and the DSIT mechanism.

In the present study, the effect of strain path reversals on DSIT was studied using an API grade X70 microalloyed steel deformed under DSIT conditions with single and multiple strain path reversals. By drawing parallel comparisons with the observations made on austenite model alloys in previous studies [25,26], insights were gained that help to understand the role played by strain path reversal on influencing the evolution of austenite grain boundaries and sub-grain boundaries and, thus, the final transformed microstructure through the DSIT process.

2. Experimental

The material used in the present study is a commercially produced low-carbon microalloyed pipeline steel (API grade X70) supplied by Tata Steel RD&T, UK. This X70 steel, with a chemical composition of 0.036C–1.56Mn–0.31Si–0.16Cr–0.16Ni–0.039Nb–0.029Al–0.014Ti–0.005Mo–0.004V–0.008P–0.006S (wt.%), was received as a hot-rolled plate with a thickness of 19 mm. The as-received plate was solution heat treated at 1250 °C for 3 h in a nitrogen atmosphere, followed by immediate water quenching to maintain the solute elements present at 1250 °C in solid solution. As shown in Fig. 1, an austenite microstructure consisting of equiaxed grains of fairly uniform diameter was achieved after the solution heat treatment. The austenite grain size measured by the mean interception length was 65.1 ± 3.7 \( \mu \)m in the rolling direction and 65.6 ± 4.3 \( \mu \)m in the transverse direction (errors represent 95% confidence limit). Solid-bar torsion specimens of 20 mm gauge length and 10 mm diameter were then machined according to the geometry described elsewhere [16].

Hot torsion tests with single and multiple strain path reversals were conducted using the servo-hydraulic Arbitrary Strain Path (ASP) testing rig at The University of Sheffield. Samples were heated by an induction method at 12 °C s\(^{-1}\) to 1250 °C, held for 2 min then cooled to the deformation temperature of 820 °C at 5 °C s\(^{-1}\). This deformation temperature is believed to be below the \( A_{c3} \) temperature of the heat-treated and undeformed X70 steel, which

![Fig. 1. (a) Optical micrograph of a uniform austenite microstructure achieved after heat treatment at 1250 °C for 3 h in a nitrogen atmosphere; (b) the position of the effective radius within the gauge section of a solid bar torsion specimen.](image-url)
is around 844 °C based on its chemical composition calculated using both JMatPro and Pandat 8.1 thermodynamic software. The $A_r$ temperature without deformation during 5 °C s$^{-1}$ cooling is believed to be around 730 °C, as indicated by dilatometry. A separate specimen was subjected to the same thermal cycle with holding at the deformation temperature for 10 s without any deformation, followed by water quenching to verify whether static phase transformation could take place at 820 °C.

Cyclical torsion tests of two and eight passes (2-pass and 8-pass tests) were conducted isothermally at a constant angular speed, which produced a constant strain rate of 1 s$^{-1}$ at the effective radius, i.e. $\sim$72.4% of the gauge radius, as shown schematically in Fig. 1(b). The concept of effective radius significantly minimizes the complexity of calculating the shear stress (hence the von Mises equivalent stress) from recorded machine torque data for torsion tests, whilst providing a basis for consistent microstructural (and textural) analysis. Detailed discussions on effective radius can be found in Refs. [26,27]. The 2-pass test consisted of forward torsion to a von Mises equivalent strain of 1.0 ($\varepsilon_{vm} = 1.0$ at the effective radius), followed by an immediate reverse torsion of $\varepsilon_{vm} = 1.0$ to a total accumulative strain of 2.0, albeit a net strain of 0. The delay due to reversing was less than 0.2 s. The 8-pass test consisted of four cycles of forward–reverse torsion, with each pass of $\varepsilon_{vm} = 0.25$, producing the same total strain of 2.0 and a net strain of 0.

After deformation, specimens were subjected to either water quenching (WQ), accelerated air cooling (AC) or slow induction coil cooling (SC). The delay between straining and quenching/cooling was less than 0.2 s. The recorded quenching rate from the deformation temperature of 820 °C to 200 °C was greater than 170 °C s$^{-1}$. The cooling rate by AC was around 15 °C s$^{-1}$, and was 0.3 °C s$^{-1}$ for SC.

Metallographic specimens of tested materials were prepared to reveal the tangential plane which is normal to the radius direction ($r$) at the effective radius and contains the axial ($Z$) and shear ($\theta$) directions of the torsion test, as outlined in Fig. 1(b). To avoid inconsistency, all microstructure observations were made along the central line of the tangential plane, as schematically illustrated in Fig. 1(b), due to a strain gradient existing on this plane [26]. Micrographs of secondary electron contrast were taken on specimens etched with 2% Nital using an FEI InspectF field emission gun scanning electron microscope (FEGSEM) at 20 kV. Electron backscattered diffraction (EBSD) maps were collected on specimens polished further by a colloidal silica suspension using an Oxford Instruments HKL Nordlys F+ camera with Channel 5 fast acquisition software in an FEI Sirion FEGSEM operating at 20 kV. A step size of 0.25 μm was used to cover typical areas of 500 μm ($Z$ direction) by 300 μm ($\theta$ direction). As the EBSD maps are commonly not fully indexed, a post-acquisition noise reduction procedure which produces consistent quantitative metallography measurements was applied [28].

### 3. Results

#### 3.1. As-quenched microstructure without deformation

The as-quenched microstructure from the X70 specimen subjected to the thermal cycle described above without any deformation is shown in Fig. 2 by (a) secondary electron contrast and (b) a coloured inverse pole figure (IPF) map. No statically transformed ferrite was observed after the 10 s holding at 820 °C. The austenite transformed into martensite after the quenching.

#### 3.2. Flow behaviour

The isothermal-corrected stress–strain curves converted from torque–angle data are shown in Fig. 3. The overall flow stress level of the 2-pass test is higher than that of
the 8-pass test. In both tests, Bauschinger effects were observed upon strain reversal. Continuous work hardening was observed during each pass of the 8-pass test. For the 2-pass test, a peak flow stress was reached around a strain of 0.8 during the forward torsion to \( \varepsilon_{\text{vm}} = 1.0 \). Afterwards, a slight dynamic softening occurred until the interruption. Immediately after the strain path reversal, a moderate drop in the flow stress (~20 MPa) was observed. The overall trend of the flow stress during the reverse torsion is to increase with accumulated total strain. However, small-amplitude fluctuations of the flow stress were also observed during the second pass.

It is known that the hot strength of ferrite is much lower than that of austenite at the same deformation temperature [29]. As the DSIT introduces new undeformed ferrite grains to the microstructure, this process could cause dynamic softening to the flow stress of the X70 steel. A recent study by Park et al. [30] demonstrated that dynamic softening in stress–strain curve indeed corresponds to the amount of dynamically transformed ferrite. The same study also showed that the tendency of DSIT increases with increasing Zener–Hollomon parameter \( Z = \dot{\varepsilon} \cdot \exp\left[\frac{Q_{\text{def}}}{RT}\right] \), where \( \dot{\varepsilon} \) is the strain rate, \( R \) is the gas constant, \( T \) is the instantaneous absolute temperature and \( Q_{\text{def}} \) is the apparent activation energy for deformation [30]. In the present work, at the current low deformation temperature of 820 °C and the relatively high strain rate of 1 s\(^{-1}\), i.e. a high \( Z \) condition, Dynamic recrystallisation (DRX) of austenite is extremely unlikely, and dynamic recovery should be limited too. Consequently, the onset of DSIT is considered to be the only dominant dynamic softening mechanism at 820 °C. Similar conclusions, i.e. DSIT causing dynamic softening, have been reached to explain the deformation behaviour during hot compression of a low-carbon microalloyed steel at 845 °C at the much slower strain rate of 0.001 s\(^{-1}\) [6]. Therefore, the dynamic softening observed during the forward torsion to a strain of 1.0 and the flow stress fluctuations occurring during the reverse torsion to a strain of 1.0 in the 2-pass test in this study are believed to be caused by DSIT as well.

The work hardening rate \( (\Theta = \partial \sigma / \partial \varepsilon) \) plotted against flow stress and strain (i.e. the Kock–Meecking plot [31]) has been widely used to determine the critical strain for DRX during hot deformation. The same method is used here to determine the critical strain for DSIT. The rationale is that the underlying principle is believed to be the same for both DSIT and DRX, as both softening mechanisms would affect the work hardening behaviour of the deformed material, which is indicated by an inflection point on the work hardening rate curves [32,33]. This method has been successfully applied to study the critical stress and strain conditions for dynamic nucleation of ferrite in steels [33]. Such plots of work hardening rate for forward torsion to \( \varepsilon_{\text{vm}} = 1.0 \) of the 2-pass test are shown in Fig. 4(a) against stress and (b) against strain. Similarly, by applying the “double-differentiation” method developed by Poliak and Jonas [32,34] for studying the initiation of DRX, the critical stress for DSIT can be determined by the stationary point \((-\partial^2(\partial \sigma / \partial \varepsilon) / \partial \sigma = 0\) on the double-differentiation curve, i.e. the derivative of work hardening rate \( (\Theta = \sigma / \varepsilon) \) with respect to flow stress \( \sigma \) as shown in Fig. 4(c). It is concluded that a critical stress of \( \sigma_{c,\text{DSIT}} = 198 \) MPa and a critical strain of \( \varepsilon_{c,\text{DSIT}} = 0.5 \) are required for the onset of DSIT under the current experiment conditions.

The fluctuations of flow stress observed during the second pass (reverse torsion of \( \varepsilon_{\text{vm}} = 1.0 \)) in the 2-pass test are believed to be a combined result from the working hardening of the remaining untransformed austenite and the DSIT ferrite, as well as from softening mechanisms, which include further DSIT, dynamic recovery of the ferrite due to its high stacking fault energy (SFE) and possible continuous DRX (CDRX) of ferrite through grain subdivision, which has been suggested by several researchers [7,35,36]. As it is generally believed that the ferrite is softer than the austenite at the same deformation temperature, the introduction of DSIT ferrite into austenite could lead to more preferential plastic deformation in the ferrite. As the deformation proceeds, the accumulation of work hardening in DSIT ferrite would, in turn, increase the macroscopic flow stress and favour strain partitioning in austenite, promoting further dynamic austenite to ferrite transformation. Work by Adachi et al. [37] using an in situ neutron diffraction method showed that deformation in the two-phase region introduces plastic strain in both the austenite and the ferrite. Therefore, further deformation could lead to the activation of some dynamic softening mechanisms in the ferrite, primarily dynamic recovery. These softening mechanisms would contribute to the decreasing of flow stress observed during the second pass of the 2-pass deformation.

### 3.3. Microstructure of deformed then quenched specimens

The microstructures of the specimens subjected to water quenching immediately after the 2-pass deformation...
(2-pass-WQ) and the 8-pass deformation (8-pass-WQ) were analysed using scanning electron microscopy (SEM) and EBSD.

3.3.1. SEM topographic contrast

The microstructure of the 2-pass-WQ test is shown in Fig. 5. It can be seen that the majority of the austenite had already transformed into very fine quasi-polygonal ferrite (QPF) grains, which are believed to be produced by the DSIT mechanism. This observation is consistent with the flow curve/work hardening rate analysis that the forward torsion strain of $\epsilon_{vm} = 1.0$ exceeded the critical strain of 0.5 for initiating the DSIT. Therefore, the deformed austenite transformed into the fine ferrite grains by the DSIT mechanism during further straining. A very small fraction of the remaining austenite transformed into lath martensite (plates in three dimensions) by a displacive transformation mechanism due to the rapid quenching after deformation. Such examples can be seen in Fig. 5 (a), indicated by the white arrows.

However, after the 8-pass, only a small amount of the DSIT ferrite grains were found, mainly at prior-austenite GBs, indicated by the white arrows in Fig. 6 (a). The interior of austenite grains remained largely untransformed during deformation, and therefore transformed into lath martensite after the subsequent quenching.

As highlighted in Fig. 6 (b), the plates are shorter and more chaotically arranged compared to the near-parallel very long and thin plates in the martensite transformed from the undeformed austenite (shown in Fig. 2). This is due to mechanical stabilization of the displacive transformation, i.e. the moving glissile transformation interface being hindered by existing dislocation substructures within the austenite matrix [38].

When the undeformed austenite decomposes by displacive phase transformations, if the material was constrained by macroscopic geometry, the shape (volume) change associated with displacive transformation would be accommodated by plastic deformations to the austenite matrix [39]. To minimize the strain energy of this shape deformation, the displacive transformation products tend to develop a morphology of thin plates. The shape deformation would also increase the dislocation density in the austenite matrix. As the displacive transformation proceeds, the dislocation density before the transformation front will continue to increase until the glissile interface is arrested as the coordinated movement of atoms is no longer sustained across these crystallographic defects. Therefore, other thin plates of the transformation product are activated repeatedly, leading to the typical microstructures of nearly parallel long plates or sheaves in martensites and bainite.

However, conditions in the heavily deformed austenite matrix are quite different. Experimental results from previous investigations on Fe–30 wt.% Ni [25] and 316L...
model austenitic alloys showed that, even with multiple strain path reversals, strong grain subdivision is still observed within the austenite grains due to the high total strain of 2.0. The large plastic deformation generated very high dislocation densities in the austenite, which are generally configured as extensive network of substructures with incidental dislocation boundaries (IDBs) delimiting ordinary cells and geometrical necessary boundaries (GNBs) separating cell blocks. Even a number of high-angle boundaries (HABs) could be created as a result of the texture/subgrain rotation grain subdivision mechanism. These GNBs, and especially the HABs, are considered to be impenetrable for moving dislocations [40]. Therefore, these dislocation boundaries would strongly interact with the shape deformation caused by the displacive transformation and the coordinated movement of the transformation front could be deterred, leading to the more chaotic morphology of martensite observed in deformed austenite.

### 3.3.2. EBSD orientation image microscopy

An IPF map and a boundary map (low-angle boundaries (LABs) > 2° as blue lines and HABs > 15° as red lines) of a selected area of the 2-pass-WQ test are shown in Fig. 7(a) and (b), respectively. The DSIT process during the 2-pass deformation produced large amounts of very fine QPF grains (2–3 μm in size), with a small number of LABs within ferrite grains, possibly due to further deformation after the DSIT ferrite was formed. The areas with a low density of HABs (red) and a high density of LABs...
in the boundary map (Fig. 7(b)) are displacive transformation products due to the quenching of the untransformed austenite. The corresponding areas on the IPF map show that these martensitic plates share similar orientations (colours), which are also indications of displacive transformation products [38]. The high level of LABs within the displacive transformation products is due to the shape deformation associated with martensitic transformation.

Similarly, the IPF and boundary maps of the as-quenched 8-pass-WQ test are given in Fig. 7 (c) and (d), respectively. As shown in Fig. 7 (c), after the 8-pass test, only coarse displacive transformation products were observed. The more granular feature of these martensite plates is believed to be due to the mechanical stabilization discussed above. It can be seen from the boundary map (Fig. 7(d)) that the density of HABs is much lower, hence the spacing of HABs is much greater compared to the 2-pass-WQ. Most of the HABs are packet boundaries between martensitic variants [41]. The very high density of LABs within each martensite packet is believed to be the result of transformation plasticity similar to that observed in the displacive products after the 2-pass test.

These microstructure features can be seen more clearly on maps produced from smaller subsets of the large EBSD maps. Such IPF coloured maps and the boundary maps are given in Fig. 8(a) and (b) for the 2-pass-WQ specimen and in (d) and (e) for the 8-pass-WQ specimen, respectively. Additionally, the local misorientation maps were produced by a 7 pixel by 7 pixel filter, with boundaries of disorientations greater than 5° excluded. They are shown in Fig. 8(c) and (f) for the transformation products from the 2-pass and 8-pass conditions, respectively.

As can be seen in Fig. 8, the QPF grains produced by the DSIT mechanism from the 2-pass test have slightly serrated GBs and a relatively low density of LABs within the ferrite grains. These features are believed to be caused by the further straining of the DSIT ferrite after they were formed when the forward strain exceeded the critical strain of 0.5. However, because of the very high SFE of ferrite, rapid ferrite dynamic recovery could occur during further deformation. This leads to relatively low variations of local orientations within the DSIT ferrite grains, as shown in Fig. 8(c).

Conversely, as the DSIT process was retarded by the repeated strain path reversals, the majority of the austenite remained deformed but untransformed after the 8-pass deformation. When water quenching was applied immediately after the deformation, the austenite transformed by a displacive transformation mechanism into martensite. The shape deformation associated with the displacive transformation generated higher levels of LABs (given in Fig. 8(f)) within the martensite packets.

Based on large-area EBSD maps, the boundary densities in terms of absolute number per unit area are plotted...
against the disorientation angle values as distribution histograms, to reveal the boundary characteristics of the microstructures after the 2-pass-WQ test and the 8-pass-WQ test. Such distribution histograms for all boundaries, including both LABs and HABs with disorientations from 2 to 62.8°, are shown in Fig. 9 (a), whilst histograms for only HABs (θ > 15°) are shown in Fig. 9(b).

Both microstructures have a similar level of LABs, although generated by different mechanisms, as discussed above. However, the levels of HAB densities between 15 and 55° are much higher for the 2-pass than for the 8-pass condition. The high densities are due to the large amount of very fine ferrite grains produced by DSIT during the 2-pass test compared to the much bigger spacing of the HABs produced by the displacive transformation after the 8-pass deformation. The peaks at around 60° disorientation are mainly due to the specific orientation relationships (OR) required for displacive phase transformations.

A grain size parameter $d_{\text{lin}}$ [28], which is the geometric mean of the linear interception lengths in horizontal and vertical directions from the EBSD maps, has been used to characterize the microstructures against predefined disorientation criteria, in this case interceptions with all boundaries having a disorientation greater than 5° and with only HABs (θ > 15°). The results are listed in Table 1, with errors representing the 95% confidence level of the measurements. The mean values of $d_{\text{lin}}$ measured against 5° disorientation are very close to each other: 2.0 ± 0.1 μm after the 2-pass-WQ test compared to 2.4 ± 0.2 μm after the 8-pass-WQ test. Nevertheless, the difference is statistically significant at $p < 0.05$. The difference between the mean linear intercept lengths of HABs (θ > 15°) is much greater – 2.8 ± 0.2 μm after the 2-pass test and 4.2 ± 0.5 μm after the 8-pass-WQ test – which means that the average HAB spacing for the latter is 150% that for the former. Overall, it can be seen from these measured mean intercept lengths that the microstructure of the 2-pass-WQ test specimen is much finer than that of the 8-pass-WQ test one.

For the as-quenched microstructure after the 2-pass-WQ test, the increase in linear interception distance is small, from 2.0 to 2.8 μm, when the minimum disorientation criteria is increased from 5 to 15°. This observation suggests that only limited deformation substructures were
developed in the very fine QPF. Conversely, for the 8-pass-WQ test, the greater increase, from 2.4 to 4.2 \( \text{\(\mu\)m} \), indicates that extensive LABs were developed due to the shape deformation of displacive transformation.

Based on the above flow stress and microstructure analysis, it can be concluded that, for the 2-pass-WQ test, the large strain \((\varepsilon_{vm} = 1.0)\) of the first pass during forward torsion exceeded the critical strain \(\varepsilon_{c,DSIT} = 0.5\) for initiating DSIT mechanism. Therefore, a large amount of very fine QPF grains were formed during the further straining of the first pass as well as during the second pass. A small fraction of remaining austenite transformed into martensite when the specimens were water-quenched after the deformation. In contrast, for the 8-pass-WQ test, multiple strain path reversals, combined with small strains for each pass \((\varepsilon_{vm} = 0.25)\), suppressed the DSIT process, even when the same amount of total accumulative strain of 2.0 was applied at the same deformation temperature. Therefore, almost no \(\gamma\) phase transformed into \(\alpha\) phase by a DSIT mechanism during deformation. The austenite subsequently decomposed to produce mainly a martensitic microstructure after the water quenching. Detailed discussions on the possible causes for this observed strain path effects on DSIT process are presented in Section 4.1.

3.4. Microstructures after fast cooling

Accelerated air cooling was applied immediately after specimens had been subjected to the 2- or 8-pass deformation, achieving an average cooling rate of 15 \(^\circ\)C s\(^{-1}\).

3.4.1. SEM topographic contrast

The microstructures after accelerated cooling for the 2-pass-deformed (2-pass-AC) and 8-pass-deformed (8-pass-AC) specimens are shown in Fig. 10(a) and (b), respectively. It can be seen that the fine QPF grains produced during the 2-pass test were retained after the fast cooling with only slightly coarsening, which was confirmed by the quantitative measurement of \(d_{lin}\) given in Table 1.

The microstructure of the 8-pass-AC specimen consists of displacive transformation products (shown in Fig. 10(b)) with much bigger lath (plate) spacing compared to that of the water-quenched one, as shown Fig. 6. The small aspect ratio (i.e. plate thickness/plate length) of the thin martensitic plates produced by water quenching was adopted to minimize the strain energy in the austenite [38]. However, during fast continuous cooling, the cooling rate \((\sim 15 \text{ C s}^{-1})\) was much slower than that of water quenching \((>170 \text{ C s}^{-1})\). Therefore, the actual transformation took place at higher temperatures, when the austenite is softer, so that higher aspect ratios of the plates can be allowed. Nevertheless, limited diffusion during the fast cooling still leads to displacive transformations.

3.4.2. EBSD orientation image microscopy

Selected area IPF maps and boundary maps of the microstructures subjected to fast cooling are shown in Fig. 11(a) and (b) for the 2-pass-AC test and in (c) and (d) for the 8-pass-AC test. Based on large-area EBSD data, distribution histograms of disorientation angles densities for these specimens are shown in Fig. 12(a) for all boundaries with disorientations from 2 to 62.8\(^\circ\) and in Fig. 12(b) for only HABs with misorientation \(>15^\circ\).

Table 1

<table>
<thead>
<tr>
<th>Test ID</th>
<th>2-pass-WQ</th>
<th>8-pass-WQ</th>
<th>2-pass-AC</th>
<th>8-pass-AC</th>
<th>2-pass-SC</th>
<th>8-pass-SC</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cooling</td>
<td>Water quenched ((&gt;170 \text{ C s}^{-1}))</td>
<td>Accelerated cooled ((\sim 15 \text{ C s}^{-1}))</td>
<td>Slow cooled ((\sim 0.3 \text{ C s}^{-1}))</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>(\theta &gt; 5^\circ)</td>
<td>(2.0 \pm 0.1 \mu\text{m})</td>
<td>(2.4 \pm 0.2 \mu\text{m})</td>
<td>(2.6 \pm 0.6 \mu\text{m})</td>
<td>(4.0 \pm 0.3 \mu\text{m})</td>
<td>(3.7 \pm 0.1 \mu\text{m})</td>
<td>(6.1 \pm 0.3 \mu\text{m})</td>
</tr>
<tr>
<td>(\theta &gt; 15^\circ)</td>
<td>(2.8 \pm 0.3 \mu\text{m})</td>
<td>(4.2 \pm 0.5 \mu\text{m})</td>
<td>(3.8 \pm 0.3 \mu\text{m})</td>
<td>(6.8 \pm 0.8 \mu\text{m})</td>
<td>(4.1 \pm 0.2 \mu\text{m})</td>
<td>(6.9 \pm 0.4 \mu\text{m})</td>
</tr>
</tbody>
</table>

Errors represent 95% confidence levels of the measurements.
boundary spacing/size parameter $d_{\text{lin}}$ measured against boundaries of disorientation angles greater than 5 and 15° are listed in Table 1.

3.5. Microstructures after slow cooling

The microstructures of the slow coil-cooled specimens after the 2-pass test (2-pass-SC) and the 8-pass test (8-pass-SC) are presented here. A linear cooling rate of 0.3 °C s$^{-1}$ was applied.

3.5.1. SEM topographic contrast

The final microstructures after both tests consist of polygonal ferrite (PF) and pearlite. The microstructure of the 2-pass-SC specimen consisted of a large fraction of fine ferrite, some coarse ferrite and some small levels of pearlite (shown in Fig. 13(a)). Furthermore, the morphology of the ferrite change to PF with fairly smooth GBs after the slow cooling compared to the QPF with serrated GBs after water quenching and fast air cooling. For the microstructure after the 8-pass-SC test (Fig. 13(b)), only large statically transformed ferrite and pearlite were found.

3.5.2. EBSD orientation image microscopy

The IPF coloured maps and boundary maps of the slow-cooled specimens are presented in Fig. 14(a) and (b) for the 2-pass-SC test and in (c) and (d) for the 8-pass-SC test. For both slow-cooled specimens, polygonal ferrite contains very few LABs. In fact, most of the LABs were found only in the pearlite microstructure. The lack of a dislocation substructure in the fine ferrite grains in the 2-pass-SC specimen (see Fig. 14(b)) suggests that extensive restoration of the DSIT ferrite took place during cooling. However, the grain growth of such DSIT ferrite seems to be very limited. Possible reasons are discussed in Section 4.2. Meanwhile, the untransformed austenite decomposed to coarse PF and pearlite by a static reconstructive phase transformation. As the static transformed PF grains were formed through nucleation and grain growth by diffusion, they were almost strain free. Therefore, a low number of subgrain dislocation boundaries were developed within static PF grains. On the other hand, as the DSIT process was retarded during the 8-pass test, the majority of the austenite remained untransformed after deformation, and subsequently decomposes to coarse PF and pearlite during the slow cooling (shown in Fig. 14(c) and (d)).

The boundary characteristics of the above microstructures are presented in Fig. 15 as boundary density–disorientation distribution histograms. Both microstructures show similar levels of LAB densities with disorientation from 2 to 10°. Furthermore, the absolute values of the LAB densities in the microstructure after the slow cooling, which are around $5.0 \times 10^4$ mm$^{-2}$, are only 1/30 of the LAB densities in the microstructures after water quenching (see Fig. 9) and accelerated air cooling (see Fig. 12), which are around $1.5 \times 10^6$ mm$^{-2}$. This is due to the effect of different cooling rates on the transformation microstructure evolution of the X70 steel.

The microstructure size parameter $d_{\text{lin}}$ measured against boundaries of disorientations greater than 5 and 15° in microstructures after the 2- and 8-pass-SC tests are given in Table 1. There are significant increases of $d_{\text{lin}}$ measured against 5° disorientation in both slow-cooled microstructures compared to the ones after accelerated cooling. This is due to the very low LAB densities within PF grains. Meanwhile, the increase in HAB spacing from AC to SC is very limited for both the 2-pass and 8-pass conditions.

4. Discussion

4.1. Effect of multiple strain path reversals on DSIT

The effect of deformation on the austenite and its role on ferrite grain refinement through conventional TMCP have
long been recognized. It was observed that ferrite nucleation occurred significantly more frequently within the deformed austenite GBs than the non-deformed austenite GBs [19,42]. In addition to increasing the total $S_v$, the primary effect of deformation is the introduction of serrations and bulges to austenite GBs which are believed to be active sites for ferrite nucleation during phase transformation [4,18]. The ferrite nucleation rate in deformed austenite could be further increased by the formation of dislocation substructures in the vicinity of austenite GBs.

Deformation of austenite also promotes the formation of intragranular planar defects such as deformation bands, which have been demonstrated to be the second most potent sites for ferrite nucleation [43]. These deformation bands in hot-deformed austenite are believed to consist of GNBs, such as dense dislocation walls and/or microbands, similar to those that were found in cold-deformed face-centred cubic (fcc) metals and alloys [44]. GNBs of high energy, i.e. higher disorientations, were found to act more effectively for ferrite nucleation during austenite decomposition [17].

Similar nucleation mechanisms have been proposed for DSIT ferrite [9]. It was found that, in the initial stage of the DSIT process, i.e. when the deformation strain just exceeded the critical strain of $\varepsilon_{c,DSIT}$, DSIT ferrite nucleated mainly on or near prior-austenite GBs [45], similar to what was observed during static transformations. A recent study on the effect of austenite grain size on DSIT
kinetics showed that smaller austenite grains, and hence a higher number of grain boundaries, accelerate the DSIT process [46]. However, the presence of $e_{c, \text{DSIT}}$ suggests that even intergranular GB nucleation requires enough strain to cause sufficient serrations and bulging of austenite GBs. During further deformation, intragranular nucleation of DSIT ferrite was activated in the interior of austenite grains [47].

There are two types of dislocation boundaries generated by deformation, namely, GNBs and IDBs. According to the power law relationship established between the average disorientation angle and the applied von Mises equivalent strain ($e_{vm}$) for both GNBs and IDBs [48], the average disorientation of GNBs increases much faster with increasing strain than that of IDBs. For example, in cold-deformed aluminium and nickel [49], the mean disorientation of GNBs increased from $\sim 1.0^\circ$ at a strain of 0.1 to above $10^\circ$ at $e_{vm} = 2.0$. However, the increases in mean disorientation for IDBs with increasing strain were found to be very small, e.g. from $\sim 0.5^\circ$ at a strain of 0.1 to only $\sim 2^\circ$ even when strained to $e_{vm} = 2.0$ [49]. Therefore, IDBs are very unlikely to function as DSIT nucleation sites. Considering that the mean disorientation of GNBs increases rapidly with strain and that there is a critical strain for DSIT, it is logical to suggest that intragranular nucleation of DSIT ferrite could only take place when certain level of...
high disorientations of GNBs is reached, e.g. $\theta > 10^\circ$, as suggested by Hodgson et al. [9].

The exact transformation mechanism of DSIT ferrite is still not fully understood. Some researchers suggest that long-range diffusion of all atoms is required, which is essentially the same as that of pro-eutectoid ferrite [50]. Others suggest a mechanism similar to massive transformation which only involves short-range diffusion across $\alpha/\gamma$ interfaces [12]. They quoted experimental evidence suggesting that there is limited diffusion of carbon during DSIT transformation as a significant number of very fine Fe$_3$C precipitations ($<20$ nm) within DSIT ferrite were found using transmission electron microscopy [23]. This strongly indicates that the carbon in DSIT ferrite is initially supersaturated, which results in the precipitation of Fe$_3$C after deformation. Moreover, a recent study [51] suggested that if the deformation temperature is well above the $T_0$ temperature, long-range diffusional strain-induced $\gamma \rightarrow \alpha$ transformation would take place; however, if the deformation temperature is below or close to $T_0$, strain-induced massive transformation will occur. These finding suggests that DSIT transformations are essentially the same as static reconstructive phase transformations; that is, allotriomorphic and idiomorphic ferrite forms above $T_0$ and massive ferrite forms below $T_0$ [52]. Of course, in the case of dynamic transformation, $T_0$ will be shifted to higher temperatures due to the deformation. Nevertheless, there is no doubt that DSIT is a form of reconstructive phase

Fig. 14. EBSD maps showing microstructures of deformed then slow-cooled ($\sim 0.3^\circ$C s$^{-1}$) X70 steel as: coloured IPF orientation maps for specimens subjected to (a) the 2-pass-SC test and (c) the 8-pass-SC test; boundary maps for specimens after (b) the 2-pass-SC test and (d) the 8-pass-SC test. (For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)
The DSIT process could be significantly influenced by the development of nucleation sites and, to a lesser extent, by growth behaviour. Therefore, the distinctive retardation of DSIT by multiple strain path reversals would not contribute to any observed difference between the two tests.
such DSIT nucleation sites, i.e. the GBs serrations and bulges as well as the generation of GNBs with high disorientation, especially HABs by grain subdivision.

Firstly, although GB serrations can develop due to the growing fluctuations of austenite GB shape with straining in both tests, the extent of GB serrations was found to be much higher after the 2-pass test than after the 8-pass test (see the examples made on a 316L stainless steel subjected to the same testing scheme in Ref. [26]). Similar to the observed retardation of DRX in 316L after the 8-pass deformation [26], the suppression of DSIT in the 8-pass test could be attributed to the following reasons. As determined by the macroscopic flow stress analysis, the critical strain of DSIT ($\varepsilon_{c,DSIT}$) is around 0.5, which is about twice the amplitude ($\varepsilon_{vm} = 0.25$) of the forward torsion in the 8-pass test. Therefore, it is likely that during each forward-torsion pass in the 8-pass test the critical strain level that could cause enough bulging of austenite GBs to facilitate the nucleation of DSIT ferrite is not reached. The subsequent reverse strain of $\varepsilon_{vm} = -0.25$ would restore the original austenite grain shape and reduce the GB area, thus eliminating further GB serration and bulging. As a result, DSIT was suppressed by the unique strain path of the 8-pass test, i.e. a combination of multiple strain reversals with small strain amplitude of each pass. Conversely, the large amplitude of forward strain $\varepsilon_{vm} = 1.0$ in the 2-pass test well exceeded the $\varepsilon_{c,DSIT}$ of 0.5. Therefore, it could introduce enough GB bulging to initiate DSIT during the forward straining.

Secondly, as austenite GBs are the preferred nucleation sites for DSIT, the amount of increment in GB areas caused by deformation would affect the DSIT process as well. According to the method developed by Zhu et al. [58], the maximum increase in the GB area introduced by the small amplitude of forward strain $\varepsilon_{vm} = 0.25$ in the 8-pass test is only $\sim 4\%$ (for details see Ref. [26]). In comparison, the large amplitude of forward strain $\varepsilon_{vm} = 1.0$ in the 2-pass test could introduce an $\sim 35\%$ increase in the austenite GB area, thus producing a much higher number of potential ferrite nucleation sites, as well as causing more extensive GB serrations and bulging. The difference between nucleation site densities caused by the different strain path histories of the two deformation schemes would contribute to the significantly different amounts of DSIT ferrite grains after the 2-pass and 8-pass deformation paths.

Finally, as shown by previous work using an Fe–30 wt.% Ni model austenite alloy [25], GNBs and IDBs could be formed through grain subdivision by both dislocation accumulation/microstructure mechanism and subgrain rotation/texture mechanism. However, the generation of GNBs with higher disorientations, especially the HABs ($\theta > 15^\circ$), could also be significantly retarded by the repeated strain path reversals in the 8-pass test. The possible reasons have already been discussed in Ref. [25]. The conclusion is again that the small amplitude of forward strain in the 8-pass test was below the threshold strain level for the activation of the subgrain rotation/texture mechanism to generate HABs [25]. Therefore, by comparing the observations made in the Fe–30 wt.% Ni model alloy and in X70 steel, it is very plausible to suggest that the generation of GNBs with high disorientations in the austenite of the X70 steel was also suppressed by multiple strain path reversals. Consequently, much less intragranular nucleation sites were available for DSIT ferrite after the 8-pass test than after the 2-pass test.

Based on the above observations, a schematic illustration of the DSIT nucleation process during monotonic deformation is proposed in Fig. 16. When the austenite is strained beyond the critical value, the first generation of DSIT ferrite starts to nucleate at a bulged area of the serrated austenite GBs which is very similar to DRX. With further straining, the second generation of DSIT ferrite starts to nucleate at the $\gamma/\alpha$ interface. Meanwhile, the disorientation of GNBs within the austenite increases to facilitate the intragranular nucleation of DSIT ferrite. More generations of DSIT ferrite will form by both intergranular and intragranular nucleation to consume the original austenite grains when the deformation proceeds.

4.2. Microstructural evolution through continuous cooling

Post-deformation cooling has been recognized as an important processing parameter to obtain a UFF microstructure [7]. A faster cooling rate could reduce the strain level required to obtain UFF after continuous cooling phase transformation [45,47]. However, it has been well recognized in previous studies that the limited coarsening and coalescence behaviours of DSIT products during post-deformation continuous cooling are instrumental in producing a UFF microstructure. The reason is attributed to the concurrent and further deformation during DSIT process. Firstly, as suggested by Beladi et al. [7], unlike the static transformation in conventionally thermomechanical

![Fig. 16. Schematic illustration of the nucleation process of DSIT ferrite during monotonic deformation: nucleation starts at bulges of serrated austenite GBs, then at austenite/ferrite interfaces and intragranular GNBs with high disorientations.](image-url)
processed austenite, which only provides a fixed number of nucleation sites, the ongoing deformation during DSIT could enable more nucleation sites in the interior of austenite grains, thus promoting extensive intragranular nucleation of ferrite, limiting further grain growth. Secondly, as suggested by Adachi et al. [37], when neighbouring ferrite nuclei share the same orientations they can easily coalesce and coarsen (Fig. 17(a)), even when the nucleation rate is high. However, the post-transformation deformation could cause the orientations of the later formed DSIT ferrite to deviate from the orientations of the ferrite grains transformed at an earlier stage (shown in Fig. 17(b)). Using an Ni–43 wt.% Cr alloy to study the orientation relationship (OR) between the fcc matrix and the body-centred cubic (bcc) precipitates, Adachi and Tsuzaki [59] demonstrated that post-transformation deformation could significantly increase the deviation angles (up to 13° for the plane parallel OR) from the ideal Kurdjumov–Sachs (KS) type OR. They attributed this to the local lattice rotation in the fcc matrix at fcc/bcc interface boundaries caused by further deformation through strain/dislocation accumulation. Similarly during DSIT processing of steels, newly formed ferrite grains favour the KS OR with respect to the already locally rotated austenite matrix. The increased orientation variation of austenite would produce neighbouring DSIT ferrite grains of different orientations, thus suppressing grain coalescence.

Very limited coarsening of DSIT products were observed in the 2-pass deformed X70 steel when the cooling rate was decreased from water quenching (170 °C s⁻¹) through accelerated air cooling (15 °C s⁻¹) to slow coil cooling (0.3 °C s⁻¹). As shown in Table 1 and Fig. 18, the corresponding HAB spacing measured against 15° disorientation increased only slightly, from 3 μm to 4 μm. These observations provide further evidence to support the argument for limited DSIT ferrite coarsening, which is unique to dynamic transformation.

Although the morphology of the transformation products after the accelerated cooling is quite different from that after the slow cooling, the overall mean HAB spacing measured against 15° disorientation in microstructures subjected to the same deformation history is not very large, as shown in Fig. 18. We speculate that there is some correlation between the HAB spacing of the transformed product and the total S_v in the deformed austenite regardless of the continuous cooling rate, which is worth further investigating.

5. Conclusions

In the present work, the effects of multiple strain path reversals on the DSIT process when X70 steel is deformed to large accumulative strains were studied by 2-pass and 8-pass cyclical torsion at temperature between the Ae₃ and Ar₃ of the steel. The microstructure evolution after DSIT and subsequent continuous cooling were also investigated. The following conclusions can be drawn:

1. When deformed to the same total accumulative strain of 2.0, deformation with multiple strain path reversals and small strain amplitude in each pass (8-pass, ε_v of 0.25 at each pass) suppressed DSIT of the X70 steel compared to deformation with a single strain path reversal and large strain amplitude in each pass (2-pass, ε_v of 1.0 at each pass), which produced extensive DSIT ferrite.

2. The critical strain ε_c,DSIT = 0.5 for initiation of DSIT during monotonic deformation was determined from the macroscopic flow stress analysis during the forward torsion of the 2-pass test using the Poliak-Jonas second derivative method, which was originally used for the study of DRX.

3. The large strain amplitude of 1.0 during forward torsion of the 2-pass test surpassed the critical strain. Therefore, extensive DSIT was generated.
On the other hand, the small strain amplitude of 0.25 in each forward pass of the 8-pass test was well below the critical strain. Therefore, DSIT process could not be initiated.

4. It is believed that the DSIT transformation mechanisms are essentially the same as the reconstructive mechanisms during static phase transformations. The retardation of DSIT by multiple strain path reversals in the 8-pass test was caused by delaying the formation of high-density ferrite nucleation sites through the suppression of austenite GBs serration and bulging, as well as by delaying the generation of deformation-induced GNBs of high disorientations. These planar defects are believed to be the most important nucleation sites for DSIT ferrite.

5. Subjected to continuous cooling after DSIT deformation, the transformation products of the 2-pass deformed specimens exhibited very limited coarsening. This is attributed to the ongoing deformation, which introduces high nucleation site density in the austenite matrix and produces orientation variations of the DSIT ferrite grains inherited from austenite.

6. As DSIT was suppressed in the 8-pass tests, the subsequent continuous cooling rate determines the transformation mechanism and the morphology of the final products, i.e. displacive transformation during accelerated cooling and reconstructive transformation during slow cooling.

From the above discussions it can be concluded that, when strain path reversals are involved in deformation to large strains, the total accumulative strain itself alone is not sufficient to determine the DSIT behaviours. The most influential factor is the relationship between the amplitude of monotonic strain and the critical strain of DSIT ($\varepsilon_{\text{c,DSIT}}$) for the steel under the current deformation conditions. If the former exceeds the latter, DSIT could be initiated. However, if strain is reversed well before the monotonic strain is required to initiate the DSIT mechanism, the strain path becomes an important issue. Large monotonic strain is required to initiate the DSIT mechanism ($>\varepsilon_{\text{c,DSIT}}$) and further deformation is needed to produce UFF microstructures ($>\varepsilon_{\text{c,UFF}}$) after continuous cooling. More capable rolling equipment is needed to provide a sufficient roll force to impose the required large reductions at a relatively low temperature. Alternatively, better rolling techniques with fewer strain path changes may also provide better grain refinement of the final products.

References