DYNAMIC SOFTENING PROCESSES IN AUSTENITIC AND DUPLEX STAINLESS STEELS DEFORMED IN HOT TORSION

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Abstract

The aim of the present study was to undertake a detailed microstructural investigation of the dynamic softening processes in both a 21Cr-19Ni-3Mo austenitic stainless steel and a 21Cr-10Ni-3Mo duplex stainless steel, containing about 60% austenite, subjected to hot deformation in torsion. Deformation was performed at 1200°C at a strain rate of 0.7 s⁻¹ to several strain levels. Quantitative optical and transmission electron microscopy together with convergent beam electron diffraction were used in the study. It was concluded that, in the single-phase steel, austenite softened via dynamic recrystallisation (DRX). The corresponding DRX grains largely nucleated along the original high-angle boundaries through the strain-induced boundary migration mechanism, complemented by (multiple) twinning. In the case of the duplex steel, DRX within the austenite was noticeably suppressed and largely restricted to the vicinity of the austenite/ferrite interface. In this region, DRX grains nucleated both through strain-induced migration of the austenite grain boundaries, complemented by (multiple) twinning, and directly from subgrains. Some fraction of the softening on the duplex steel flow curve was also attributed to large-scale subgrain coalescence within the dynamically recovered austenite matrix. Ferrite present within the duplex microstructure softened via “extended” dynamic recovery, characterised by a gradual increase in misorientations between neighbouring subgrains with strain.

1. INTRODUCTION

It has been widely accepted that austenite in steels, due to its relatively low stacking fault energy, tends to soften during hot deformation by “discontinuous” dynamic recrystallisation (DRX) which involves nucleation and growth of new recrystallised grains and often occurs in conjunction with dynamic recovery (DRV) [1]. Conversely, ferrite having a comparatively high stacking fault energy has a tendency to soften by intense dynamic recovery (DRV) during hot working. There has been extensive work done to establish the hot deformation conditions under which austenite softens by either DRV or DRX in various steels and alloys [1]. The above work, however, was mostly done using an analysis of flow curves in combination with optical microscopy and there has been lack of detailed information about substructure and crystallography of dynamically recovered and especially dynamically recrystallised austenite in the literature. Development of DRX within austenite in duplex stainless steels is particularly interesting because of limited availability of pre-existing austenite/austenite boundaries that are known to serve as major DRX nucleation sites in single-phase austenite [1,2]. There are three possible softening mechanisms of ferrite: DRV, DRX and “extended” DRV (sometimes called “continuous” DRX) characterised by gradual increase in misorientations between neighbouring subgrains with strain [3]. Softening processes within ferrite, despite their complexity, have attracted rather limited attention.
The aim of the present study was to undertake a detailed microstructural investigation of the softening processes in both a 21Cr-19Ni-3Mo austenitic stainless steel (steel A) and a 21Cr-10Ni-3Mo austenite/ferrite duplex stainless steel (steel D), subjected to hot torsion at 1200°C.

2. EXPERIMENTAL PROCEDURES

The chemical compositions of the steels studied are given in Table 1. Volume fraction of the austenite, present within the microstructure of steel D, was estimated to be about 60%. Specimens with a gauge length of 50 mm and a diameter of 6 mm were subjected to deformation in torsion at a strain rate of 0.7 s⁻¹ at 1200°C followed by rapid quenching for retaining the deformation microstructure. The equivalent Von Mises stress (σ) and strain (ε) were calculated by standard methods. Torsional deformation was performed either continuously to the point of cracking or was interrupted at several strain levels selected for the various parts of the flow curves. These strain levels were approximately 0.3, 0.7, 1.4, and 2.5 for the single-phase steel A and 0.1, 0.5, 0.9, and 1.3 for the duplex steel D. Metallographic examination was carried out on tangential sections at a depth of 1 mm under the surface using quantitative optical microscopy and transmission electron microscopy (TEM). Convergent beam electron diffraction (CBED) was used to study local crystallographic orientations and misorientations.

Table 1.
Chemical composition of the steels used (wt%).

<table>
<thead>
<tr>
<th>Steel</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>P</th>
<th>S</th>
<th>Cu</th>
<th>Cr</th>
<th>Ni</th>
<th>Mo</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>0.020</td>
<td>0.99</td>
<td>0.55</td>
<td>0.023</td>
<td>0.016</td>
<td>0.16</td>
<td>21.54</td>
<td>18.61</td>
<td>2.93</td>
</tr>
<tr>
<td>D</td>
<td>0.020</td>
<td>0.96</td>
<td>0.51</td>
<td>0.023</td>
<td>0.017</td>
<td>0.11</td>
<td>21.15</td>
<td>10.05</td>
<td>3.02</td>
</tr>
</tbody>
</table>

3. RESULTS

3.1 Austenite Softening

Figures 1a and 2a show the continuous flow curves of both steels A and D. Both of the curves were characterised by a flow stress peak at a strain of about 0.6, in the vicinity of which a small amount of DRX nuclei was observed in the austenite. A gradual increase in DRX volume fraction was accompanied by a parallel decrease in flow stress. However, a steady-state regime, that would be characterised by a constant flow stress and a completely dynamically recrystallised microstructure, was far from being reached at the point of cracking in either of the steels studied. In the single-phase steel A, cracking occurred at a strain of approximately 2.5 giving rise to a DRX volume fraction of about 30%. The corresponding decrease in flow stress, relative to the peak stress value, was approximately 28%. In the case of the duplex steel D, at the point of cracking (a strain of about 1.3), only approximately 10% of the austenite was found by optical microscopy to be dynamically recrystallised. However, the corresponding drop in flow stress reached about 27%. Mean DRX grain size values remained approximately constant in both steels during straining, irrespective of strain, which is good correspondence with the published data [1,2]. These values were about 30 µm and 6 µm for steel A and D respectively.
From an inspection of the optical micrograph shown in Fig. 1b it is clear that, in the single-phase steel A, DRX grains nucleated predominantly in triple junctions and through bulging of the original austenite grain boundaries that became progressively serrated with increasing strain. Such a DRX nucleation mechanism, commonly observed in single-phase austenite, has been called “strain-induced boundary migration” (SIBM) [2]. Some intragranular DRX nucleation was also observed to take place both through SIBM of the distorted pre-existing twin boundaries and on various inhomogeneities of the deformation microstructure. In the duplex steel D, some of the DRX grains were also observed to be formed through SIBM of the original austenite grain boundaries (Fig. 2b). In this case, however, this nucleation mechanism was frequently detected in the vicinity of the austenite/ferrite (A/F) interface, as illustrated in Fig. 3. This figure also shows that SIBM was often complemented by the formation of (multiple) twins at the migrating boundary, which was observed in both the steels studied.

![Figure 1](image1.png)

**Figure 1.** (a) Continuous flow curve of the single-phase steel A; (b) optical micrograph of the corresponding microstructure at a strain of 1.4.

![Figure 2](image2.png)

**Figure 2.** (a) Continuous flow curve of the duplex steel D; (b) optical micrograph of the corresponding microstructure at a strain of 1.3 (austenite is white and ferrite is grey).

Apart from SIBM utilising the pre-existing austenite high-angle boundaries, some DRX nuclei were observed to be formed by recovery processes from subgrains in steel D. This nucleation mechanism predominantly took place along the A/F interface. Figure 4a shows an austenite region situated next to the A/F interface and containing such a DRX nucleus. The
Figure 3. SIBM in the vicinity of the A/F interface, complemented by twinning, observed within the duplex steel D at a strain of 1.3: (a) TEM bright-field image; (b) corresponding schematic interpretation (A and F denote austenite and ferrite respectively, GB indicates the migrating austenite grain boundary).

nucleus was separated from the neighbouring well-recovered subgrains by mostly large-angle boundary facets. It was generally observed that the subgrains situated in the vicinity of the A/F interface were characterised by significantly larger misorientations across their boundaries compared to those lying further away from the interface, which made them suitable candidates to become potential DRX nuclei. Furthermore, the subgrain misorientations had a tendency to accumulate strongly with distance within the A/F interface region (Fig. 4b), as opposed to to the rest of the austenite matrix where the misorientation gradients accumulated across consecutive subboundaries were significantly smaller (Fig. 4c).
In both the steels studied, TEM analysis revealed the presence of the localised austenite regions, containing dissolving subboundaries accompanied by an increased density of individual dislocations released from the corresponding dislocation networks (Fig. 5). This seems to be a manifestation of the well-known subgrain coalescence mechanism [2] but, in this case, often involving unusually large numbers of subgrains. Such a large-scale subgrain coalescence was found to be particularly widespread in the duplex steel D where, at the point of cracking, the volume fraction of coalesced regions was well around 20%. These regions largely remained an integral part of the dynamically recovered austenite matrix and their boundaries were frequently not well defined. Alternatively, the above mechanism was frequently observed to clear the dislocation substructure from the interior of both the pre-existing annealing twins and the hierarchical fragments delineated by larger-angle boundaries, that were originally subdivided by low-angle subboundaries. In the former case, “recrystallised” regions delineated by high-angle boundaries were thus created in analogy to DRX grains.

![Figure 5. TEM bright-field micrograph showing a region of coalesced austenite subgrains observed within the duplex steel D at a strain of 0.9.](image)

### 3.2 Ferrite Softening

A careful examination of the ferrite in the microstructure of the duplex steel D, using optical microscopy, revealed that the original ferrite grains were gradually subdivided during straining into subgrains, delineated by boundaries characterised by differing etching response (see Fig. 2b). TEM analysis showed that the subgrains were characterised by a roughly equiaxed morphology mostly with a relatively low dislocation density in the interior (Fig. 6a). A detailed investigation of the misorientation characteristics of subgrain boundaries was undertaken using the specimen deformed to the point of cracking (a strain of about 1.3). About 400 boundaries were analysed and the distribution of misorientation angles obtained is presented in Fig. 6b. The mean subgrain size and the mean misorientation angle were about 7.6 µm and 4.8° respectively for this strain level. The subgrain boundaries were observed to form a complex network composed of a mix of small- and large-angle boundaries. It is necessary to note that the boundaries characterised by misorientations above 20° were never observed to be part of this network.
4. DISCUSSION

4.1 Austenite Softening Mechanisms

The results obtained indicate that a significant number of DRX austenite grains nucleated through the SIBM process in both the steels studied, utilising the pre-existing austenite grain or distorted twin boundaries, which is a commonly observed DRX nucleation mechanism [2]. In the present case, this mechanism was frequently complemented by the formation of (multiple) twins at the migrating boundary, which was also reported in [4,5]. It has been suggested that the reason for multiple twinning might be a tendency of the migrating boundary to either decrease its energy [4] or increase its mobility [6]. In the duplex steel D, however, only part of the DRX austenite grains was found to nucleate via the SIBM mechanism. The limited availability of the original austenite grain boundaries in the duplex microstructure does not seem to be the only reason for this behaviour. Even though the nucleation sites at these boundaries were not exhausted, there was a large amount of new DRX grains nucleated next to the A/F interfaces directly from subgrains.

This could be expected from the observed characteristics of the corresponding deformation substructure. The quite homogeneous subgrain structure with very small misorientations across the subboundaries, found further away from the A/F interfaces, does not seem to have always provided sufficient accumulated energy gradients across the austenite grain boundaries for SIBM to occur. In contrast, the regions close to the A/F interfaces were observed to have a comparatively significantly larger stored deformation energy, as a result of accommodation of sliding and incompatibility strains, manifested by large subgrain misorientations that strongly cumulated with distance. Thus, new large-angle boundaries delineating dislocation-free regions could have been created in the vicinity of the A/F interface through recovery processes and their subsequent migration, in particular along this interface, might have led to the formation of DRX grains. This type of DRX nucleation has also been reported in [7].

The very low misorientations between neighbouring subgrains within the austenite, in conjunction with the high deformation temperature used, seem to have promoted localised subgrain coalescence, in particular in the duplex steel D. The regions of coalesced subgrains, provided that they were created in the actual process of straining (dynamically), could be expected to contribute to the decrease in flow stress in a similar manner as DRX grains. These regions do appear to have largely been created dynamically, as the torsion specimens were quenched extremely quickly after deformation (within a fraction of second) and, thus, the
deformation microstructure could be expected to remain essentially retained. Thus, the formation of a large volume fraction of dynamically coalesced subgrains, observed within the austenite in the duplex steel studied, might have contributed substantially to the softening on the corresponding flow curve. This could account for the discrepancy between the relative decrease in flow stress (approximately 27%) and the corresponding DRX volume fraction (only about 10%), observed in the above steel at the point of cracking.

The formation of the observed regions of coalesced subgrains does not seem to have been a manifestation of the so called “repolygonisation” mechanism [8], characterised by repeated formation and disintegration of dislocation walls, that has been suggested to accompany dynamic recovery processes. The coalesced regions observed in the present study were clearly formed through the simultaneous destabilisation of a large number of subboundaries. Such a process locally led to the formation of almost dislocation-free “recrystallised” areas without the involvement of noticeable migration of large-angle boundaries. It has been well established that subgrains may rotate by diffusional processes until adjacent crystals attain a similar orientation, thus eliminating the pre-existing low-angle subboundaries [2]. The driving force for this process arises from a reduction in the total boundary energy. Nevertheless, the concrete reason for the occurrence of the large-scale subgrain coalescence during hot torsion, observed under the present experimental conditions, still remains to be clarified. This process might possibly represent an alternative mode of austenite dynamic softening to the discontinuous DRX in a microstructure, characterised by limited availability of the pre-existing austenite high-angle boundaries serving as prominent DRX nucleation sites, such as the austenite/ferrite matrix in steel D studied. Thus, as a result of the dynamic large-scale subgrain coalescence, a significant level of softening on the flow curve might be achieved despite a very low fraction of DRX grains observed by optical microscopy.

4.2 Ferrite Softening Mechanism

From the results obtained in the present study it is clear that no evidence, suggesting the occurrence of discontinuous DRX, was found within the ferrite in the duplex steel D under the deformation conditions used. A detailed TEM analysis clearly revealed that the dislocation substructure within the original ferrite grains was well recovered and dislocations were mostly arranged into perfect planar walls characterised by a wide range of misorientation angles that were as large as 20°. The process leading to the formation of such a pronounced hierarchy of dislocation boundaries might be classified as an extension of “classical” DRV that is usually connected with the mechanism of repolygonisation [8], characterised by the repeated formation and disintegration of dislocation walls. Repolygonisation tends to maintain a low-angle character of dislocation boundaries and also to produce an approximately constant size and equiaxed shape of subgrains within the steady state flow regime.

The observed dislocation substructure was clearly a result of the operation of intense DRV, accompanied by a gradual increase in the mean misorientation angle across subgrain boundaries during straining. It may be more appropriate to term this process “extended DRV” rather than “continuous DRX” often used in the literature [3]. The mechanism of extended DRV suggested in [9] seems to predict well the character of the arrangement of subgrain boundaries observed in the present study. According to this concept, only less than one third of glide dislocations are expected to be annihilated when they meet each other with the rest being stored in dislocation networks, a majority of which have uncompensated stress fields. Migration of low-angle dislocation walls under the action of both the mutual attraction forces and the external stress field leads to their merging with a change in misorientation. Consequently, gradual build up of misorientations might take place in different parts of the
initial grains through low-mobility boundaries. The processes of migration and merging of low-angle dislocation walls tend to create a complex equilibrium network of small- and large-angle boundaries [9].

5. CONCLUSIONS

A detailed microstructural investigation of the dynamic softening processes within both austenite and ferrite has been undertaken in a 21Cr-19Ni-3Mo single-phase and a 21Cr-10Ni-3Mo duplex stainless steels, the latter containing about 60% austenite, subjected to hot torsion at 1200°C at a strain rate of 0.7 s\(^{-1}\). Dynamic recrystallisation was observed to develop within the austenite during straining in both steels, concurrently with dynamic recovery. In the single-phase steel, the nucleation of dynamically-recrystallised grains largely occurred through the strain-induced migration of the pre-existing austenite high-angle boundaries, complemented by (multiple) twinning. In the duplex steel, the above DRX nucleation mechanism was frequently found to take place in the vicinity of the austenite/ferrite interface, where it was accompanied by the formation of DRX nuclei directly from subgrains, through dynamic recovery processes. In both steels, but in the duplex steel in particular, some fraction of dynamic softening might be also attributed to large-scale subgrain coalescence. The softening process within the ferrite in the duplex steel has been classified as “extended” dynamic recovery, characterised by a gradual increase in misorientations between neighbouring subgrains with strain.

REFERENCES