

# Effect of $\delta$ -ferrite co-existence on hot deformation and recrystallization of austenite

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**Abstract** This work evaluates the effect of co-existence of a large volume fraction of  $\delta$ -ferrite on the hot deformation and dynamic recrystallization (DRX) of austenite using comparative hot torsion tests on AISI 304 austenitic and 2205 duplex stainless steels. The comparison was performed under similar deformation conditions (i.e. temperature and strain rate) and also under similar Zener-Hollomon,  $Z$ , values. The torsion data were combined with electron backscatter diffraction (EBSD) analysis to study the microstructure development. The results imply a considerable difference between DRX mechanisms, austenite grain sizes and also DRX kinetics of two steels. Whereas austenitic stainless steel shows the start of DRX at very low strains and then development of that microstructure based on the necklace structure, the DRX phenomena in the austenite phase of duplex structure does not proceed to a very high fraction. Also, the DRX kinetics in the austenitic steel are much higher than the austenite phase of the duplex steel. The results suggest that at a similar deformation condition the DRX grain size of austenitic steel is almost three times larger than the DRX grains of austenite phase in duplex steel. Similarly, the ratio of DRX grain size in the austenitic to the duplex structure at the same  $Z$  values is about 1.5.

## Introduction

Hot deformation and dynamic recrystallization (DRX) of austenite has been studied in different steels and specifically in austenitic stainless steels as model alloys [1–4]. The effect of different parameters such as temperature, strain rate, chemical composition and initial grain size on DRX has also been examined in numerous studies [3, 5–11]. The most common understanding of DRX in austenite is that its progress is usually by serration and bulging of pre-existing grain boundaries and through the formation of a necklace structure [11]. Therefore, the initial grain boundaries and their frequency as effective nucleation sites have pronounced effects on the DRX of austenite. By increasing the grain boundary frequency (by decreasing the initial grain size), the nucleation sites of DRX increases and recrystallization accelerates [12]. However, the investigation of a single phase material with a very low number of initial grain boundaries has been hindered by the inability to obtain large grain microstructure during conventional thermomechanical processes. An alternative approach is to use a duplex (austenitic-ferritic) stainless steel. As the austenite/austenite grain boundaries in the duplex structure are very limited, this type of steel can be used as a model alloy for the above reasons. This selection also provides an extra opportunity to study the effect of a second phase on the DRX of austenite. Earlier studies [13, 14] have revealed the strong influence of second phases on the DRX mechanism, kinetics and microstructural development of austenite in plain carbon steels. This effect is particularly obvious when austenite co-exists with a large volume fraction of second phase (e.g.  $\delta$ -ferrite) in a duplex structure. Several different studies have shown the significant effect of this austenite-ferrite co-existence on the hot deformation and microstructural

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development in duplex stainless steel [15–19]. However, there is still a lack of knowledge concerning the DRX mechanism, DRX kinetics, and, specifically, the alternative source for nucleation sites of new DRX grains in the austenite phase (as the frequency of austenite internal boundaries is very low).

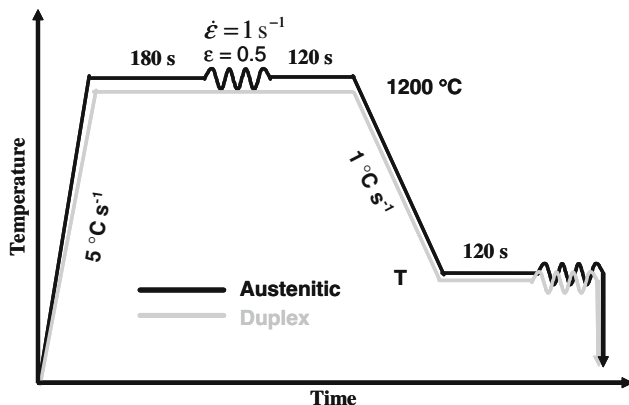
The present study aims to clarify the role of initial grain boundaries and the second phase on the hot deformation and DRX of austenite. In this regard, a comparison has been made between the DRX process in an austenitic stainless steel and in the austenite phase of a duplex stainless steel using hot torsion tests and electron backscatter diffraction (EBSD) analysis.

### Experiments

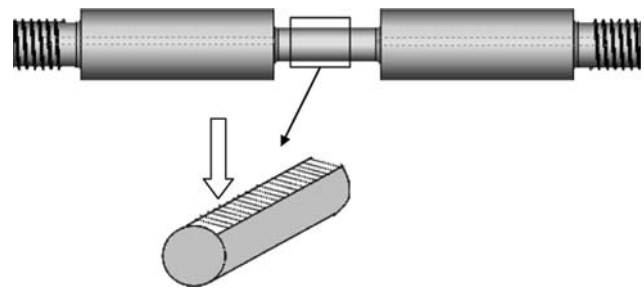
An AISI Type 304 austenitic and a 2205 duplex stainless steel with chemical compositions given in Table 1 were used in this investigation. Torsion samples with a gauge length of 20 mm and a diameter of 6.7 mm were machined from the as-hot rolled bars. The details of the hot torsion equipment and experimental methods are described elsewhere [20]. An homogenizing treatment at 1,200 °C was performed either by a roughing process (for 304 steel) or just by soaking for 300 s (for 2205 steel). The samples were then cooled to the deformation temperatures and held for 120 s before deformation. Hot deformation was performed at different strain rates and then followed by a rapid quench to room temperature (Fig. 1). The duplex steel was

**Table 1** Chemical composition of steels used in the current study (wt.%)

Alloy	C	Mn	Si	S	P	Ni	Cr	Mo	Cu	N
304	0.02	1.6	0.7	0.01	0.03	8.2	18.5	0.1	0.8	–
2205	0.03	2.0	1.0	0.02	0.03	5.5	22	3.0	0.6	0.1



**Fig. 1** Schematic diagram representing the hot torsion tests



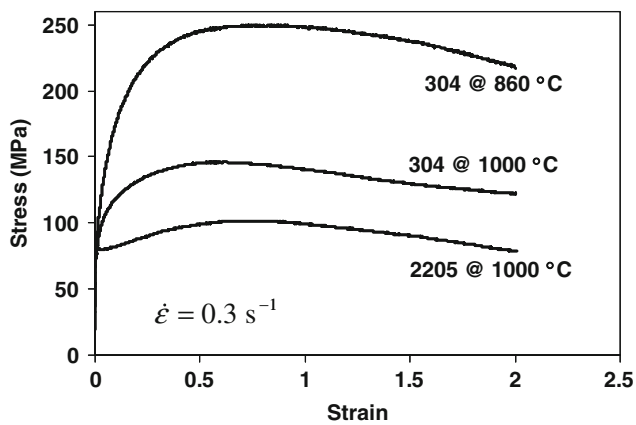
**Fig. 2** A schematic diagram from the torsion samples (the arrow shows the direction of polishing the surface (tangential section) and viewing the microstructure)

tested at 1,000 °C and a strain rate of 0.3 s<sup>-1</sup>. The Type 304 steel was tested both at the same strain rate and temperature conditions as duplex steel or at a similar Zener-Hollomon, Z, parameter (i.e. 860 °C and 0.3 s<sup>-1</sup>). In this regard, the values of 400 and 450 kJ/mol were considered for the activation energy of the AISI 304 and 2205 steels, respectively, based on the data reported in [21].

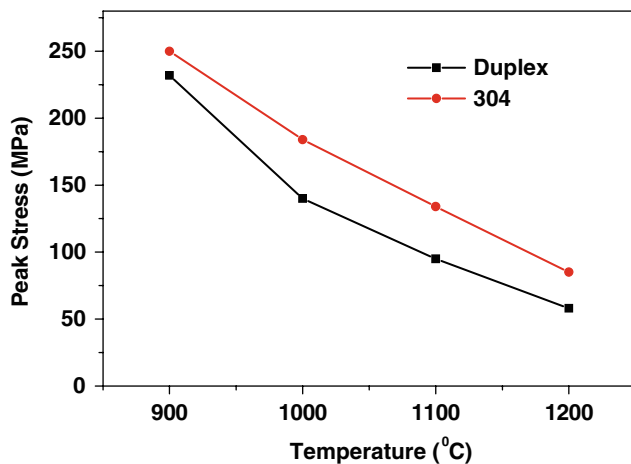
Conventional metallographic procedures were used to prepare tangential sections from the torsion sample gauge length at a depth of ~100 μm below the gauge surface of the quenched samples (Fig. 2). These polished sections were subsequently analysed by EBSD to characterize the microstructure. The EBSD analyses were performed at an accelerating voltage of 20 kV, a working distance of 25 mm, and an aperture size of 60 μm. EBSD maps were obtained using a step size of 0.5 μm, and HKL technology channel 5 software. The volume fraction of recrystallized grains and their mean size were measured from the EBSD maps using the point counting technique and linear intercept method, respectively [22].

### Results and discussion

An illustrative example of flow curves of both steels under different deformation conditions are presented in Fig. 3, which compares the flow curves under the similar deformation condition and similar Z values. The most evident difference between these flow curves is their characteristic shape. The austenitic steel showed a typical shape of DRX, i.e. increasing the flow stress to a peak followed by decreasing to a steady-state value, whereas the duplex steel showed a different behaviour. In this steel, the peak is broadened and a steady-state flow stress was also never observed after the softening stage. In fact, the stress decreased continuously to the rupture of the sample. However, the similarity between the stress at the rupture and the initial yield stress indicates the occurrence of significant softening in this material. The flow curves of duplex steel also showed different shapes compared with



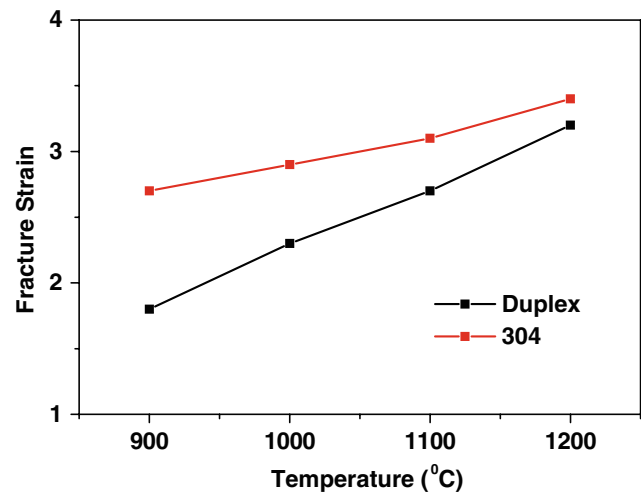
**Fig. 3** Flow curves of samples deformed under different deformation conditions



**Fig. 4** Peak stress of austenitic (304) and duplex stainless steels as functions of deformation temperature at a constant strain rate of  $0.3 \text{ s}^{-1}$

the flow curve of ferritic stainless steels cited in the literature [23–26]. The flow curve shape of duplex steel indicates that DRX of the austenite is not the only reason (or is a very minor reason) for such softening in the duplex steel.

In comparison with the AISI 304 steel, the peak stress,  $\sigma_p$ , showed a lower value for the duplex stainless steel (Fig. 4) in all deformation conditions. However, the difference decreased slightly at lower temperatures, which can be ascribed to the increase in the austenite volume fraction within the duplex structure. Also, the fracture strain of duplex samples were much lower than that for the austenitic steel (Fig. 5), indicating lower ductility in duplex structure. Such reduced ductility in the duplex structure can be a direct result of inhomogeneity in the stress and strain distribution within the duplex steel during hot working. This inhomogeneity in the strain distribution can originate from the difference in the ferrite and austenite strength,



**Fig. 5** Fracture strain of austenitic (304) and duplex stainless steels as functions of deformation temperature at a constant strain rate of  $0.3 \text{ s}^{-1}$

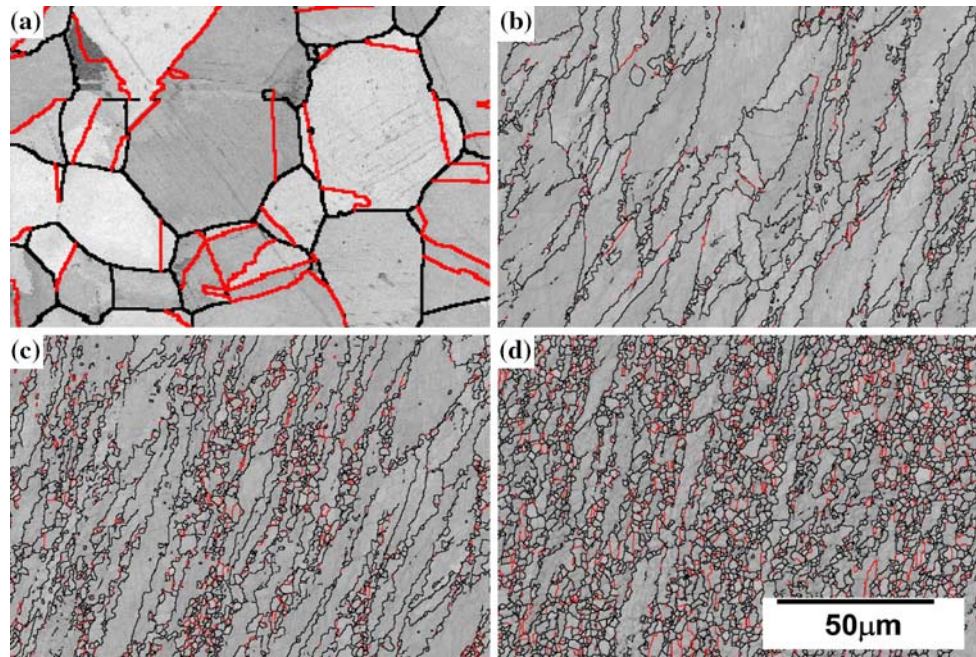
different restoration processes in these phases, and also the inhomogeneity in the distribution of austenite stringers (or layers) in the ferrite matrix. The lower strength of ferrite compared to that of the austenite can cause more accommodation of deformation in this phase, and, therefore, faster restoration processes (recovery and recrystallization). This can lead to an inhomogeneity in the duplex structure and a decrease in the ductility of the steel. The microscopic observations of different duplex steels have shown that the cracking always initiates at  $\delta/\gamma$  interfaces as a result of high stress concentrations [27].

It is well known that the deformation temperature and strain rate (both of which can be combined through the  $Z$  parameter), are the most important variables affecting the hot deformation processes and the final microstructure of a deformed and recrystallized sample [6]. Figures 6 and 7 show the initial and deformed microstructures of both the austenitic and duplex steels under similar  $Z$  conditions. The initial microstructure (i.e. after homogenization and cooling to the deformation temperature) of the AISI 304 austenitic steel (Fig. 6a) consisted of equiaxed grains with an average size of  $\sim 35 \mu\text{m}$ . All boundaries are high angle, and more than 45% have a twin relationship. In contrast, the initial microstructure of the duplex structure (Fig. 7a) consisted of similar volume fractions of austenite and ferrite distributed in successive layers or stringers. This shows that the austenite phase is fully recrystallized and most of the high-angle grain boundaries within this phase ( $>80\%$ ) have a twin relationship. In addition, no low-angle grain boundaries (LAGB's) were observed.

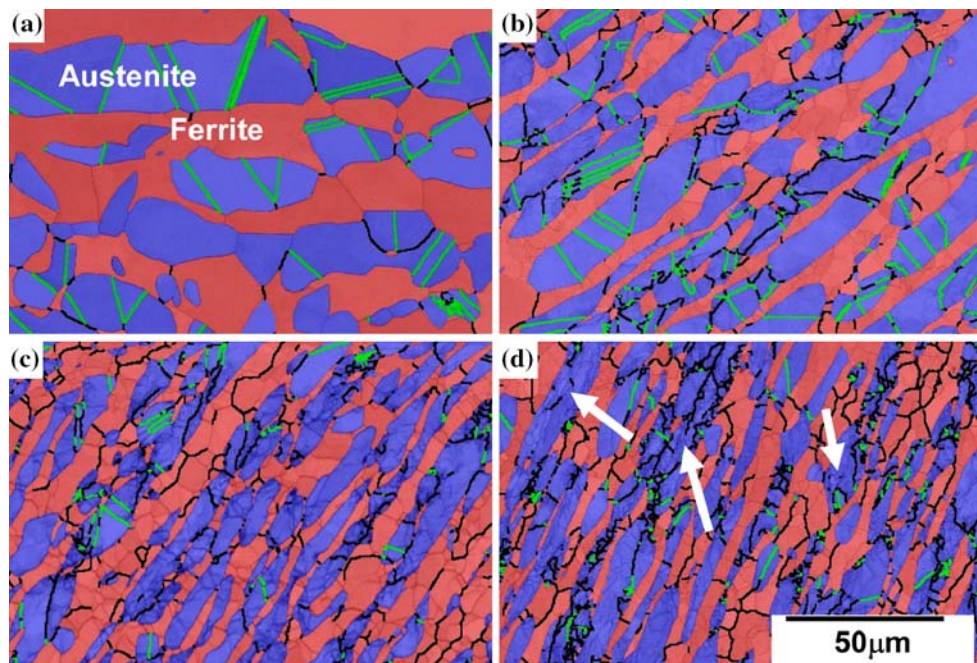
Deformation of these initial microstructures under similar  $Z$  conditions resulted in very different microstructures. The microstructural change in the austenitic steel commenced with the elongation of original grains at low



**Fig. 6** EBSD map of austenitic steel deformed at 860 °C and a strain rate of  $0.3 \text{ s}^{-1}$  to different strains of (a) Zero, (b) 1.0, (c) 1.5 and (d) 2.0 High angle and twin boundaries have been shown by black and white (red) lines, respectively



**Fig. 7** EBSD map of duplex steel deformed at 1,000 °C and a strain rate of  $0.3 \text{ s}^{-1}$  to different strains of (a) Zero, (b) 0.5, (c) 1.0 and (d) 1.5 High angle and twin boundaries have been shown by black and white (green) lines, respectively



strains, serration of their boundaries and then nucleation of new DRX grains, much smaller than the initial grains. After a strain of 1.0 (Fig. 6b) all the initial grain boundaries were serrated and some were covered by new DRX grains. The elongation of the grains and serration of their boundaries increased with increasing strain (Fig. 6c). Most of these serrated boundaries were covered by DRX grains at a strain of 1.5. The volume fraction of small DRX grains increased with increasing strain; at a strain of 2.0 most of the initial grains (more than 75%) were replaced by new

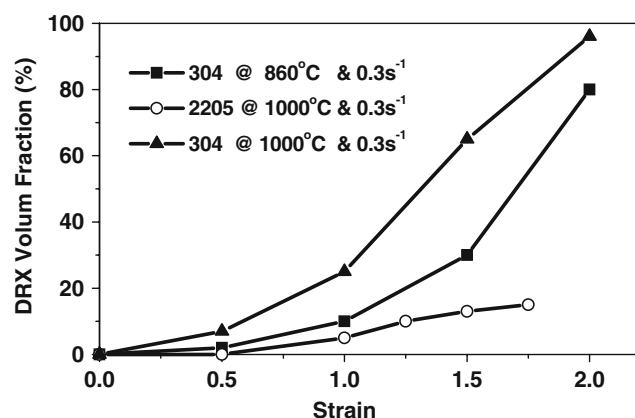
DRX grains. It appears that in the austenitic stainless steel, under the present deformation condition, most of the DRX grains were formed by serration and bulging of the initial grain boundaries. This mechanism (i.e. bulging) has been known as the most important DRX mechanism in materials with low-stacking fault energy and is usually termed discontinuous or conventional DRX [28, 29].

A different microstructure was developed in the austenite phase of duplex steel. At a low strain of 0.5 (Fig. 7b) all of the austenite and ferrite layers were elongated and

rotated towards the deformation (torsion) axis, but no obvious change was observed in the overall microstructure and substructure of the austenite. Strain partitioning and accommodation of the strain in the softer phase of ferrite is likely to be the main reason for this limited effect of deformation on the austenite in the duplex structure. By increasing the strain to 1.0 (Fig. 7c), very limited serrations were observed in the austenite/austenite boundaries and also no new austenite grains were observed. However, a very dense substructure was formed, indicating load transfer from the ferrite to the austenite. At a higher strain of 1.5, which is close to the fracture strain of this material, a limited number of new DRX grains were formed on the austenite/austenite boundaries with more new grains forming on the austenite/ferrite phase boundaries (Fig. 7d). There was a very low level of serration of boundaries in this steel, indicating the formation of new DRX grains by a different mechanism from bulging of initial grain boundaries. The formation and coalescence of subgrains (i.e. continuous DRX) could be an alternative mechanism for formation of DRX grains in this phase.

The EBSD maps of this steel show that the misorientation of some segments of the substructure network increased during deformation and changed to HAGB's (arrows in Fig. 7d). These segments may reflect an intermediate stage for formation of a complete grain by coalescence of subgrains. This increase in the misorientation of low-angle grain boundaries (LAGB's) inside the deformed grains is consistent with the continuous DRX model proposed by Gourdet and Montheillet [30]. Therefore, it can be concluded that in the absence (or for a limited proportion) of initial grain boundaries, a change from conventional to continuous DRX occurs.

The microstructural measurements also showed a considerable difference between the DRX kinetics for both steels under similar Z conditions (Fig. 8). The progress of



**Fig. 8** DRX volume fraction as a function of strain for both steels under different deformation conditions

DRX in the austenitic stainless steel was always much faster than for the duplex steel under similar deformation condition and similar Z values. As the strain rate, temperature and the initial grain size are the variables that affect the DRX kinetics (i.e. the Avrami exponent), the decrease in the frequency of internal boundaries of austenite phase in the duplex structure could be the most important reason for such differences.

As the activation energy of duplex steel (which is a mix of two phases with different activation energies) is different from the activation energy of austenite, the comparison of DRX of a given phase (i.e. austenite) under two different conditions may not be appropriate. In this case, comparison under the same deformation conditions (i.e. temperature and strain rate) may be more relevant. In this regards, the microstructure of AISI 304 austenitic stainless steel was compared with duplex steel at the same deformation condition (i.e. 1,000 °C and 0.3 s<sup>-1</sup>). The result showed that the DRX grain size (Table 2) and also the DRX volume fraction (Fig. 8) in the austenitic stainless steel are much higher than that of the austenite phase of the duplex steel. Table 2 shows that for a similar deformation condition, the DRX grain size in the austenitic steel is much larger (almost three times) than that of the austenite phase in the duplex steel. Also, for a similar Z value, the DRX grain size of the austenitic steel is larger than the duplex steel. This indicates the effect of other parameters (besides the temperature and strain rate) on DRX kinetics and the microstructure evolution of austenite in these two steels. One possible difference is in the amount of strain in both austenite phases due to strain partitioning with accommodation of less strain in the austenite of the duplex steel. However, while strain can affect the DRX kinetics, many studies, e.g. [12, 31], have shown no effect of strain on DRX grain size. The difference in the DRX mechanisms (e.g. continuous or discontinuous) may create the different DRX microstructures in these two steels. A difference between the grain size of a discontinuously recrystallized microstructure in the austenitic steel (evolved based on the serration and local migration of high-angle grain boundaries) and a continuously recrystallized structure in the duplex steel (evolved based on the coalescence of subgrains) is reasonable. However, it has been shown that when continuous DRX is involved, the migration rate of

**Table 2** DRX grain size of duplex and 304 austenitic stainless steels under different deformation conditions

	Duplex <sup>a</sup> 1,000 °C & 0.3 s <sup>-1</sup>	304 1,000 °C & 0.3 s <sup>-1</sup>	304 <sup>a</sup> 860 °C & 0.3 s <sup>-1</sup>
DRX grain size (μm)	3.2	11	4.5

<sup>a</sup> Similar Z values



high-angle grain boundaries (i.e. growth of DRX nuclei) is much lower than that of discontinuous DRX condition [30, 32]. This means that the growth of DRX grains is limited when continuous DRX is dominant, and the grain size becomes finer. Therefore, it can be concluded that a change in the DRX mechanism of austenite from discontinuous to continuous in the presence of a large fraction of ferrite, is an important reason for the different DRX microstructures and grain sizes. However, the role of second phase (i.e. ferrite) on the DRX phenomenon of austenite and strain partitioning between two phases can also be considered as an important reason for such difference in DRX grain size and DRX kinetics in two steels.

## Conclusions

The results of hot deformation tests under various conditions showed a considerable difference in restoration process of the austenite phase in a duplex stainless steel compared to that for an austenitic stainless steel. The difference in the frequency of initial austenite grain boundaries as well as presence of the second phase (in the duplex steel) were the most important reasons for observed differences in the restoration processes. Under both similar deformation conditions and similar Z values, the DRX fraction in the austenitic steel was much higher than for the duplex steel. Also, under a similar deformation condition the DRX grain size in austenitic steel was almost three times larger than the DRX grain size in austenite phase of a duplex structure. It is suggested that the conventional DRX (i.e. bulging of initial grain boundaries) and continuous DRX (i.e. coalescence of subboundaries) are the major restoration mechanisms in austenitic and duplex steels, respectively. This difference in the DRX mechanism indicates that the growth of DRX nuclei in the austenitic steel is more likely in contrast to the duplex steel.

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