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EBSD Investigation of Friction Stir Welded Duplex Stainless Steel

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Abstract**—**Electron back-scattered diffraction was used to follow the evolution of microstructure from the base metal to the stir zone (SZ) in a duplex stainless steel subjected to friction stir welding. In the stir zone (SZ), a continuous dynamic recrystallization (CDRX) was evidenced for ferrite, while it was suggested that a static recrystallization together with CDRX may occur for austenite. It was found that ferrite and austenite grains in the SZ take a typical shear texture of bcc and fcc materials respectively.

Keywords**—**Friction stir welding, Dynamic recrystallization, Electron backscattering diffraction (EBSD), Duplex stainless steel

I. Introduction

IT has been shown that friction stir welding (FSW) can alleviate most of the problems caused by the fusion welding processes in duplex stainless steels (DSS) [1]. This motivation has influenced some researchers to deal with the microstructure and mechanical properties of friction stir welded DSS [2], [3]. Sato et al. [2] conducted a detailed examination of the microstructure and the mechanical properties in the friction stir welding of SAF 2507 super duplex stainless steel. They found that FSW significantly refined the ferrite (α) and austenite (γ) phases in the SZ. Based on the obtained results, they concluded that fine grains of both α and γ increased hardness and strength in the SZ. Recently, using a WC-based tool, Saeid et al. [3] investigated the effect of welding speed on the microstructure and mechanical properties of SZ in SAF 2205 DSS. They reported that increasing the welding speed decreased the size of the α and γ grains, and hence, improved the mean hardness value and the tensile strength in the SZ.

The above-mentioned efforts have yielded some significant knowledge on microstructure, and mechanical properties in friction stir welded DSS, but relatively little attention have been paid to details of grain structure formation. Especially, recrystallization phenomena of two constituent phases have not been systematically investigated. For example, Sato et al. [2] and Saeid et al. [3] supposed, according to previous findings about conventional deformation, annealing, hot working, and superplastic processing of DSS, that discontinuous dynamic recrystallization of austenite and continuous dynamic recrystallization of ferrite simultaneously occur in the SZ. These findings were developed from experiments involving conditions of approximately uniform straining and isothermal deformation conditions. However, during FSW, material experiences large stresses and high temperatures under conditions that include rapid transients and steep gradients in strain, strain rate and temperature [4]. Therefore, the potential contributions of the steep gradients in strain, strain rate and temperature that are characteristic of FSW could not be reflected by conventional deformation methods, and it is imperative to gain better understanding and control of developed grain structure in these circumstances. The recent emphasis on production of fine and desirable microstructures, without the risk of grain growth in weld zone of duplex and microduplex stainless steels [5], has also accentuated the need for a systematic study to provide insight into mechanisms by which grain structure forms. The results of such a study in DSS, as a microstructurally less complicated material, could be used to predict the microstructural changes of hypo-eutectoid carbon steels during FSW. Since in these carbon steels γ-transus temperature is relatively low and under the condition of welding the peak temperature in the SZ generally exceeds A1 temperature. As a result, FSW is performed in α+γ two-phase region, and γ to α phase transformation during cooling cycle of weld hinders the microstructural evidences, which are affected directly from FSW. Thus, the history of microstructural changes occurred in two-phase region is difficult to trace by a postmortem microstructural analysis, such as that performed by Fujii et al. [6]. Accordingly, in this paper, we concentrate on EBSD results to investigate the microstructural and textural evolution of DSS.

II. Experimental Procedure

The base material was a SAF 2205 duplex stainless steel, which was supplied in a form of rolled and annealed (at 1050 ◦C for 1 h) plate. This plate was cut into a specimen with dimensions of 300mm×100mm×2mm and then subjected to friction stir welding in a bead-on-plate configuration. Welding was conducted along the rolling direction of the plate at a welding speed of 100 mm/min and a rotational speed of 800 rpm. Details about welding parameters, used tool and machine have been given in a previous paper [3]. Microstructural

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examination was undertaken on a weld transverse cross-section

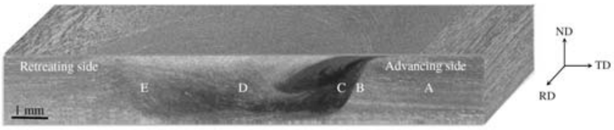

Fig. 1 Optical macrograph of SAF 2205 after FSW. Regions A and B lie in the BM and TMAZ respectively. Region C, D and E are located in the advancing side, center and retreating side of the SZ.

by optical microscopy (OM) and the EBSD technique. OM sample was electrolytically etched in a solution of 56 g KOH in 100 mL distilled water at 3 V for 30 s. EBSD specimen was mechanically ground, prepolished using 1 and 0.25 μm diamond pastes and finally electropolished using a solution consisting of 700 mL ethanol, 120 mL distilled water, 100 mL glycerol, and 80 mL perchloric acid at a temperature of approximately 25 ◦C and a voltage of 35 V for 10 s. EBSD pattern acquisition and analyses were performed on a TSL EBSD system on a JSM-6400 JEOL scanning electron microscope (SEM), operated at 20 kV. Orientation mapping was performed on a triangular scanning grid with step sizes of 0.1 - 0.5 μm.

III. Results and Discussion

A low-magnification overview of the friction stir welded area is shown in Fig. 1. Three distinct zones can be readily distinguished: a stir zone, the base metal (BM) and a narrow transition region of thermomechanically affected zone (TMAZ).

The EBSD data corresponding to locations A-C in the cross section centerline of Fig. 1 are shown in Figs. 2 and 3 in the form of phase maps, and misorientation angle histograms, respectively.

EBSD analyses revealed that the BM contained 48.5% γ phase comprising of equiaxed grains with an average size of 5 μm, embedded in the α matrix with the average grain size of 7 μm (Fig. 2a). A detailed analysis of the misorientation angle/axis pairs across the γ/γ boundaries (Fig. 3a) revealed that about 63% of the high-angle boundaries (HABs) displayed the first-order twin coincidence-site lattice (CSL) orientation relationship of ∑3. The presence of such a high quantity of twin boundaries in the microstructure demonstrates the formation of twins during recrystallization and grain growth of the starting material [7]. Furthermore, the amount of low-angle boundaries (LABs) with misorientation angles (θ) between 2◦ and 15◦ is very low in γ, while the α phase in Fig. 3d contains considerable amounts of LABs.

In the TMAZ grains are markedly bent by the plastic deformation into the direction inclined to the TMAZ/SZ boundary (Fig. 2b). As a result of such plastic deformation, the pre-existing annealing twins within the γ were rotated away from the ideal CSL orientation relationship and correspondingly, originally straight coherent twin boundaries converted to curved, general HABs. As indicated in Fig. 3b, the portion of first-order twin boundaries decreased to about 8%. Such twin boundary distortions seem to result from complex interactions of twin boundaries with slip dislocations [8].

The misorientation distribution of Fig. 3b exhibits a higher fraction of LABs for γ than that of the BM. In the case of α, the fraction of LABs also increases up to 43% (Fig. 3e) and the mean misorientation decreases to about 23◦.

Fig. 2c shows that a fine and equiaxed grain structure formed in the SZ. The fraction of γ twin boundaries in this area is increased when compared with TMAZ (Fig. 3c). However, the frequency of twin boundaries in the SZ is much lower than that observed in the base material, suggesting that the intense deformation associated with FSW broke up the initial annealed microstructure of the base material. On the other hand, the number of LABs in α and γ decreases noticeably from TMAZ to the SZ with corresponding increases in the fraction of HABs (Fig. 3c and f), which leads to an increase of the mean misorientation angle and widening of the misorientation spectra in the SZ.

The FSW process consists of a rapidly rotating tool with a shoulder that exerts a strong shearing deformation on its plane of action. This shear deformation prefers to rotate the

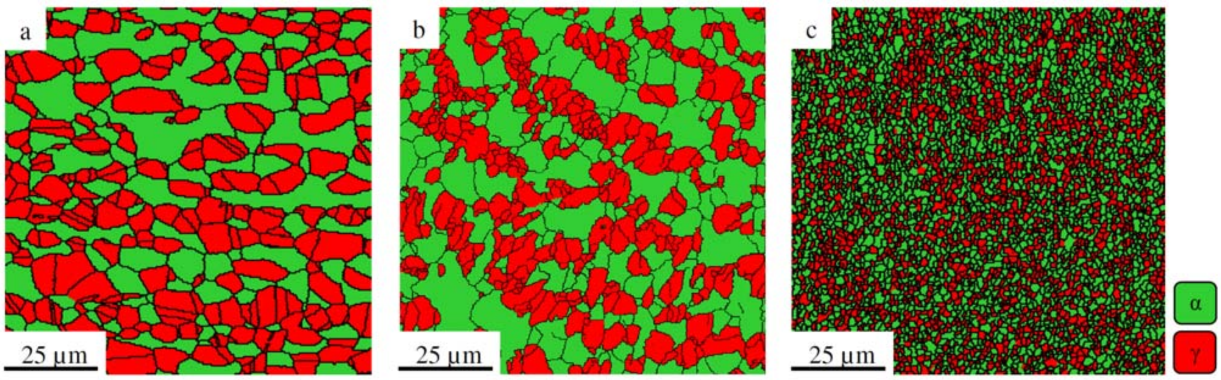

Fig. 2 Phase maps of weld showing various microstructural zones in SAF 2205: (a) BM, (b) TMAZ and (c) SZ.

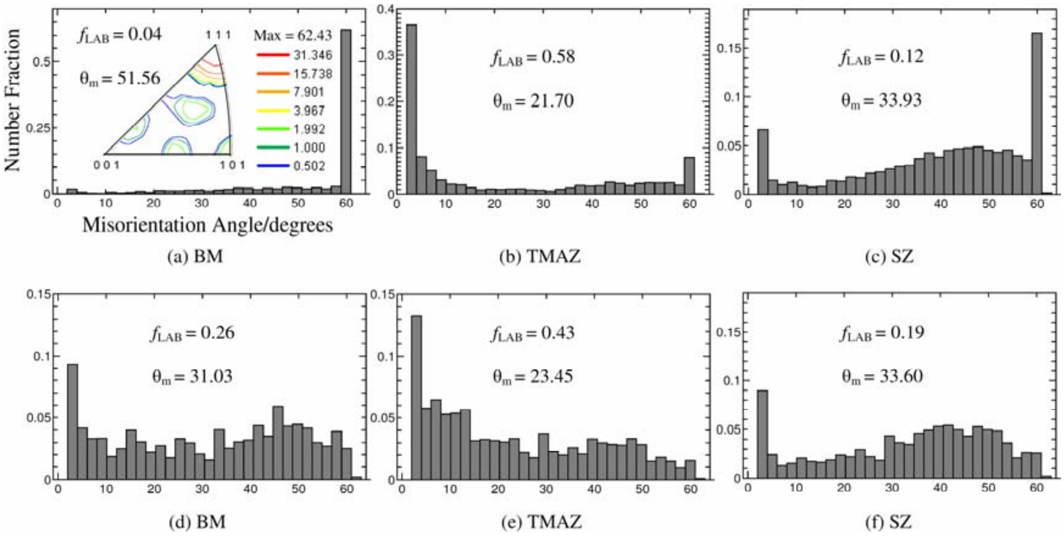

Fig. 3 Misorientation angle histogram for γ/γ (a - c) and α/α boundaries (d - f), within the BM, TMAZ, and SZ. f_{LAB} and θ_m indicate the fraction of LABs and mean misorientation angle, respectively

crystallite lattice so that the plane of easy glide lies in the shearing plane. From this point, examination of the texture may provide valuable information on deformation mechanism and resultant grain structure. Results of such texture analyses within SZ are presented in Fig. 4 in the form of {111} and {110} pole figures of γ and α, respectively. It is suggested that the texture developed in the SZ can be defined in terms of the shear plane normal (SPN) and shear direction (SD) aligned approximately perpendicular to and tangential with either cone surface of the pin or the flow lines in the stir zone, respectively [4], [9], [10]. Therefore, the pole figures of Fig. 4 are in the recalculated form, which are obtained from the orientation distribution function (ODF) and rotated around certain axes to illustrate the textural data in the simple shear reference coordination. The detail procedure of such rotations is described by Mironov et al. [9].

Fig. 4 reveals that α and γ phases have the typical simple shear texture, just like the texture produced by deformation of bcc and fcc metals during torsion testing [11]-[13]. In addition to deformation textures, the γ in the SZ showed the presence of cube texture. Inspection of experimental data shows that continuous dynamic recrystallization (CDRX) occurs in the α phase within the SZ. This mechanism of microstructure evolution consists of two sequential processes [14]:

(i) The formation of three-dimensional arrays of LABs;

(ii) The gradual conversion of LABs into HABs.

In the first process, LABs are continuously formed by dynamic recovery during deformation by rearrangement of accumulating lattice dislocations. The observed enhancement of LABs fractions of both α and γ in the TMAZ can be attributed to the occurrence of such mechanism [9]. The second process is characterized by the progressive increase of LABs misorientation through the absorption of mobile dislocations in the pre-existing boundaries, which leads to the transformation of LABs into HABs. These processes result in the formation of individual segments of HAB that appear independent of original HABs. The change of a nucleus into a grain occurs by the conversion of all its LAB facets into HABs. The above mechanism is generally associated with the presence of deformation texture [15].

The gradual increase in the mean misorientation angle and widening the misorientation spectra from the TMAZ to the SZ, and retention of a shear-type deformation texture within the SZ appear to be consistent with the occurrence of the CDRX in the α phase.

While the CDRX can account for α, it is difficult to understand the formation of γ grains by this mechanism. Indeed, the presence of cube component suggests that alternative recrystallization mechanisms may also occur in the γ during FSW of duplex stainless steel. Some previous studies [16], [17] have shown that static recrystallization (SRX) often produces texture components having cube texture, in the type 304 stainless steel. Furthermore, SRX justifies, to some extent, why the fraction of γ twin boundaries in the SZ is higher than TMAZ. Therefore, it is conceivable that the dominant orientations of γ in the SZ, would be produced by CDRX together with SRX behind the pin tool at elevated temperatures produced by the shoulder. This scenario implies that shoulder has an important role in the formation of cube component during SRX, in spite of its negligible effect on the plastic flow of studied areas in the SZ.

IV. Conclusion

In summary, the evolution of microstructure of duplex stainless steel from the base metal to the SZ was investigated. In the SZ, crystallographic texture evolution of ferrite phase was consistent with the simple shear texture characteristics reported for individual bcc metals. However, austenite phase

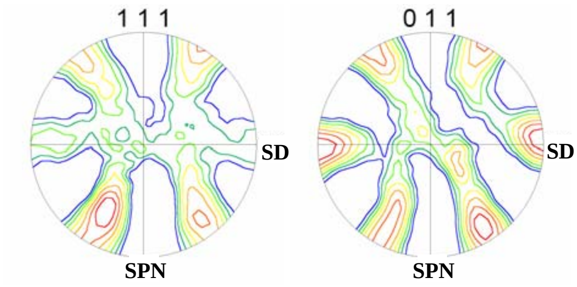

Fig. 4 {111} and {011} pole figures of austenite and ferrite, respectively, in the SZ

pole figures comprised of the superposition of fcc shear and cube components.

Based on the obtained results, it was suggested that CDRX of α occurrs in the SZ, while a complex mechanism, which may be consisting of CDRX and SRX, is responsible for γ fine grain formation.

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