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Microstructure evolution during friction-stir welding of AZ31 magnesium alloy

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ABSTRACT

The microstructure evolution of a typical hexagonal close-packed (HCP) material (AZ31 magnesium alloy) during friction stir welding was studied in a wide range of welding temperatures. In all cases, the grain structure development was found to be significantly influenced by the formation of a very strong $\{0001\}\langle uwtv\rangle$ B-fiber texture. Due to limitations imposed by this texture as well as by symmetry of the HCP crystal structure, an extensive lowering of grain-boundary misorientation was found to occur during deformation.

1. Introduction

Friction stir welding (FSW) is well recognized as a very practical technology having great industrial potential. However, this technique also represents an interest from a purely academic standpoint. During FSW, materials undergo very large strains at high temperatures and a relatively high strain rate. The materials behavior under such extreme deformation conditions has not been well studied. Thus, microstructure and texture observations in FSW may improve our fundamental understanding of materials.

To date, grain-structure development during FSW is best studied in face-centered cubic (FCC) metals. Structural behavior of these materials has been found to be principally influenced by stacking fault energy (SFE). If SFE is high, microstructure evolution is often reported to be governed by *continuous* recrystallization (e.g. [1–4]), i.e. a process which typically involves extensive formation of deformation-induced boundaries and their gradual transformation into high-angle boundaries. If SFE is low, the grain structure development is usually dominated by nucleation of recrystallization nuclei and their subsequent growth (e.g. [5–7]), i.e. *discontinuous* recrystallization. In metals with intermediate SFE (e.g. copper), a *transition* from continuous to discontinuous recrystallization with increasing welding temperature has been observed recently [8]. All these trends are believed to be closely related to variation of dislocation mobility with SFE, thus being in line with our current understanding of materials behavior.

On the other hand, the structural response of metals with a *hexagonal* crystal structure is less clear. A limited number of slip systems available in these materials as well as their propensity to mechanical twinning may significantly affect the microstructure evolution.

Among hexagonal metals, magnesium alloys are most feasible for FSW and therefore their microstructure has been most studied so far. It has been established that FSW of these materials often leads to significant grain refinement [9–18]. The microstructure in the stir zone is typically dominated by nearly equiaxed grains [10,18–20] and contains almost no twins [21], i.e. it seems to be recrystallized in appearance. On the other hand, it may also contain a relatively high dislocation density [10,21]. The stir zone material is usually characterized by the development of a very strong basal texture [10–12,14,15,17–32] and a specific misorientation-angle distribution with a sharp peak near 30° [15,22,27,31,32].

It should be emphasized that the above-mentioned research efforts were primarily focused on characterization of the *final* microstructure (or texture) which evolved in the stir zone. However, relatively little attention was paid to *how* the microstructure actually develops during welding and which basic *mechanisms* control this process. The present work was undertaken in an attempt to fill this gap in our knowledge.

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2. Material and experimental procedures

The material used in the present investigation was a commercial AZ31 magnesium alloy with a nominal chemical composition of Mg–3.0Al–1.0Zn (wt.%). This is a typical and comparatively simple hexagonal alloy whose deformation behavior has been well documented. The base material was produced by extrusion at 350 °C followed by annealing for 2 h at 300 °C. It was supplied as 4-mm thick sheets.

The as-received material was friction-stir welded in a bead-on-plate configuration. To examine microstructure evolution at different welding temperatures, FSW was performed in a wide range of tool rotational speeds, from 300 to 3000 rpm, whereas the tool travel speed was kept constant at 200 mm/min. The welding tool was fabricated from a tool steel and consisted of a concave shoulder 15 mm in diameter and an M5 threaded cylindrical pin 3.7 mm in length[1]. In all cases, the tool tilting angle was 3° and the tool plunge depth was controlled to be ~3.5 mm. The principal directions of welding geometry are denoted throughout as welding direction (WD), transverse direction (TD), and normal direction (ND).

The welding temperature was measured by K-type thermocouples placed directly on the weld seam[2]. They were typically destroyed by the passing tool[3] and the peak recorded temperature was defined as the welding temperature. As expected, the temperature increased with the tool rotational speed from $\sim 0.57T_m$ at 300 rpm to $\sim 0.85T_m$ at 2000 rpm, where T_m is the melting point (Fig. 1a). The welding temperature, however, tended to saturate at this level and did not change substantially up to the tool rotational speed of 3000 rpm (Fig. 1a). Nearly the same temperature behavior during FSW of AZ31 has been reported by Chang et al. [19]. The observed temperature saturation is thought to be related to reduction of the friction coefficient at high temperatures, as suggested by Mishra et al. [33].

The increase in welding temperature with tool rotational speed led to material softening and a respective decrease in tool load (Fig. 1b). As shown in Section 3.4, this substantially affected the microstructure evolution. Moreover, the welding at high temperatures was also characterized by relatively high stability of the tool load (Fig. 1b), thus presumably reflecting a more stable character of material flow.

The microstructure observations were mainly performed by electron-backscatter diffraction (EBSD). A suitable surface finish was obtained by electro-polishing in a commercial AC2 Struers polishing solution at ~10 °C (ice bath) with an applied potential of 40 V. High-resolution EBSD analysis was conducted using a Hitachi S-4300SE and JEOL JSM-7800F field-emission gun scanning electron microscopes equipped with a TSL OIM™ EBSD system. To ensure reliability of EBSD data, all small grains comprising three or fewer pixels were automatically removed from EBSD maps using the grain-dilation option of the TSL software. To eliminate spurious boundaries caused by orientation noise, a lower limit boundary misorientation cut-off of 2° was used. A 15° criterion was employed to differentiate low-angle boundaries (LABs) versus high-angle boundaries (HABs).

Because the evolved microstructures sometimes exhibited a complex mixture of LABs and HABs, confusion can arise with regard to the definition of grains. To avoid this ambiguity, the term "grain" in the present work was applied to denote a crystallite bordered by a *continuous* HAB perimeter.

[1] A sketch of the tool is shown in Supplementary Fig. S1.

[2] The layout of the thermocouples is shown in Supplementary Fig. S2a.

[3] To evaluate a complete weld thermal cycle, thermocouples were also placed beneath the welded sheets. Typical measured temperature profiles are shown in Supplementary Fig. S2b.

3. Results and discussion

3.1. Low-magnification overview of welds

Optical images of the friction stir welds obtained at different tool rotational speeds are shown in Fig. 2. In all cases, the retreating side is on the left and the advancing side is on the right.

The welds produced at the tool rotational speed ⩽1000 rpm contained a tunnel-type defect at the weld root (Fig. 2a and b). This observation is thought to be related to the relatively unstable character of material flow at low welding temperatures, as mentioned above. Remarkably, the stir zone of the low-temperature welds closely resembled the cylindrical shape of the tool pin (Fig. 2a). This perhaps indicated a predominance of the pin-induced material flow. An increase in tool rotational speed eliminated the welding defects and gave rise to an elliptical shaped stir zone (Fig. 2b–d). This probably reflected a change of character of material flow with welding temperature.

To check this idea, sample-scale EBSD maps were obtained from several welds, as shown in Fig. 3. In these maps, individual grains are colored according to their crystallographic orientations relative to the ND[4]. Therefore, the red-colored area in the upper parts of the welds indicates an alignment of the (0001) basal planes with tool shoulder surface, i.e. a prevalence of shoulder-induced material flow. It is seen that this area enlarged with the tool rotation speed, i.e. the contribution of the shoulder to global material flow increased with welding temperature.

Also of interest in Fig. 3 is that the texture distribution in the 1000-rpm weld seems to be less regular than that in higher-temperature welds. This effect was particularly pronounced near welding defects. This observation is also believed to be indicative of instability of material flow at low welding temperatures.

3.2. Grain structure and texture in stir zone

Attempting to avoid the effect of complicated material flow in the top and bottom part of the stir zone, microstructure examinations were focused in its central (i.e. nugget-) area[5]. The selected portions of EBSD grain-boundary maps obtained in different welds are summarized in Fig. 4. In these maps, LABs and HABs are depicted as red and black lines, respectively.

In terms of morphology, the grain structures which evolved in each weld were broadly similar to each other. In all cases, the grains were often not completely delineated by a continuous HAB perimeter and therefore the microstructure was difficult to describe in conventional terms of grains. Instead, it was rather represented by an irregular mixture of LABs and HABs. The origin of this structure is discussed in Section 3.3.1. In agreement with previous studies [21,23,31,32], the microstructures contained almost no twins.

As expected, the microstructure coarsened with welding temperature and this effect was found to be more pronounced at temperatures exceeding $\sim 0.65T_m$ (Fig. 5a).

In all cases, a development of {0001} $\langle uwtv \rangle$ B-fiber texture was revealed (Fig. 6). The formation of this texture was first reported by Park et al. [30] and subsequently confirmed in numerous works [10–12,14,15,17–29,31,32]. According to Beausir et al. [34], the B-fiber texture is attributable to the prevalence of the basal slip, as is normally expected for magnesium alloys. It should be emphasized that the evolved textures were strong – the peak

[4] Here and hereafter, a reader is referred to on-line version of this paper to see figures in color.

[5] The microstructures were found to be reasonably homogeneous in stir zones of each weld. A typical microstructure distribution is shown in Supplementary Fig. S3.