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Mechanical properties' modification of large load friction stir welded AZ31B Mg alloy joint

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ABSTRACT

Friction stir welded (FSW) Mg alloys usually exhibit an undesirable combination of strength and elongation due to its strong texture develops in the weld. Thus, large load FSW associated with an extremely low welding speed and rotation rate were applied to an AZ31B Mg alloy to modify the microstructure, the texture and the mechanical properties of the joint. The twin structure in the weld provided adequate barriers for dislocation motion for strengthening and created more local sites for nucleating and accommodating dislocations, thereby elevating ductility and strain hardening of the weld. This work also provided a simple and effective method to enhance the strength of an FSW Mg joint without ductility loss.

1. Introduction

The most important characteristic of Friction stir welding (FSW) is the grain refinement in the stir zone (SZ), which improves the alloys' strength and ductility [1]. Unfortunately, the degradation of strength in Mg alloys is serious after FSW [2–4]. The reason is that a strong crystallographic texture develops in the SZ due to the Mg alloys' limited independent slip systems and the shear stress from the rotating tool [3]. It has been commonly reported that the (0001) basal planes of most grains in the SZ are approximately parallel to the surface of pin tools [4]. For this type of texture, the SZ is a soft region in transverse tensile tests, where the grains are favorably oriented for basal slip. Therefore, significant efforts are needed to randomize the strong texture caused by FSW, such as post-welding deformation [4,5], post-welding tension associate with subsequent annealing [6] and double-sided FSW technique [7–9]. However, the two- or three-step processing method and double-side tool are not well-suited for practical applications, and the annealing process is not easily controlled due to the strong driving force of recrystallization after severe plastic deformation. Therefore, a simple processing technology should be developed to improve the texture, which, in turn, to modify the mechanical properties of the FSW Mg alloys joint. In the present work, a novel large load FSW (LL-FSW) with extremely low welding speed and rotation rate was designed to the joint of an AZ31B Mg alloy. The purpose of this research is to obtain a SZ with reduced texture intensity and a welded joint with excellent combination of strength and ductility.

2. Experiments

AZ31B Mg alloy plates with dimensions of $200^{L} \times 100^{W} \times 3^{T}$ mm^3 were butt-welded in the rolling direction using a load-control FSW machine. The WC-Co based FSW tool had a shoulder with a diameter of 15 mm and a right screw threaded pin with a height of 2.7 mm and diameter of 6 mm. Threaded pin was usually adopted in the FSW of Mg alloys to cause severe material flow and reduce the texture intensity in the weld [7–9]. The welding process was conducted at the rotation speed of 20 rpm and the welding speed of 10 mm/min with the axial load was set as 2 t. The temperature history during welding was measured using a K-type thermocouple at the bottom of workpiece interface. The SZ of LL-FSW showed a peak temperature of 168 ± 11 °C. After the LL-FSW, the microstructure was examined under optical microscopy (OM) on the cross-section perpendicular to the welding direction. The texture of the SZ center and thermo-mechanically affected zone (TMAZ) on the cross-section perpendicular to the welding direction were examined via the electronic backscattering diffraction (EBSD) technique using a TSL OIM™ system. The EBSD scanning step size was set as 100 nm. The average grain size was calculated using

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the linear intercept method. Transmission electron microscopy (TEM) was applied to examine the substructures in the SZ. The Vickers hardness distribution was measured under a load of 100 g for a dwell time of 15 s along the centerline of the cross-section with an interval of 0.5 mm. Two tensile specimens were produced to evaluate the mechanical properties of the welded joint according to the standard of JIS Z2201(13B) [10]. The transverse tensile specimens covering various regions of the FSW joints with a gauge size of $50^L \times 12.5^W \times 2.7^T$ mm^3 were machined with the tensile axis perpendicular to the welding direction. The specimens containing only the SZ were also machined with a gauge size of $12^L \times 3^W \times 2^T$ mm^3 and the tensile axis parallel to the welding direction to evaluate the tensile strength of the SZ. The tensile tests were carried out using an Instron-type testing machine at a cross-head speed of 1 mm/min at room temperature. The principal directions of this work were denoted by the welding direction (WD), transverse direction (TD) and normal direction (ND).

3. Results and discussion

Fig. 1(a) shows the cross section of the weld. Defects such as cracks and porosity were not observed, indicating that the AZ31B Mg alloy was successfully welded using LL-FSW. The hardness increased from the base material region toward the SZ, and no heat-affected zone (low-hardness zone) was observed (Fig. 1(b)). The SZ showed an enhanced hardness of 114 HV, much higher than that of the base material (74 HV). For the tensile specimen containing the base material and welded joint, the specimen fractured at the base material, indicating the ultimate tensile strength of the base material is about 148 MPa. This means that the whole welded joint including the SZ and the TMAZ was highly strengthened compared with that of the base material (Fig. 1(c)). For the tensile axis parallel to the WD, the true stress–strain curve together with the strain hardening rate (θ) vs. true strain curve of the SZ are given in Fig. 1(d). The SZ exhibited excellent strain hardening capacity, resulting an ultimate tensile strength of 292 MPa was achieved. According to the Considère's criterion, the value of uniform elongation (elongation before plastic instability, i.e. when $\theta \geq 1$) was calculated to be 26%, which was highly enhanced compared with that of the previous studies [4–9].

An orientation color map taken from the SZ center is presented in Fig. 2(a). The SZ has an ultra-fined grain structure with a mean grain size of 0.5 μm, which highly refined compared with that of the base material of 12 μm. As visible in the misorientation distribution histogram as shown in Fig. 2(b), there were two prominent peaks, which was quite different from the random distribution of the hexagonal close-packed crystallites as indicated by the black dotted line. The first was located at 2–5°, indicating a dislocation substructure. Considering the lower temperature within the LL-FSW and the annealing effect just after stirring, the so-called post-annealing effect was reduced and the amount of substructure or dislocations was retained in the SZ. The second peak was at about 86° because the basic symmetry of the hexagonal close-packed crystals included limited slip systems, making twinning an important deformation mechanism. Twinning on the pyramidal plane leads to a small amount of expansion along the c-axis, with the basal plane in the twin being reoriented through 86°. The magnified region in Fig. 2(a) displays that a twin is marked with three black hexagons to show their orientations in three dimensions. The texture component in the SZ center was significantly different compared with that of the conventional FSW [4–9] and base material of (0001) plane tends to be parallel to the TD-WD plane. The (0001) pole figure not only presents the <0001>||WD texture component but also the <0001>||ND component; consequently, the texture intensity is comparatively small. A possible reason for this change was that as the deformation temperature decreased, there was a tendency for {10–12}<10–11> twinning to occur. Such twinning would reorient the {0001} texture about 86° back towards the ND.

A careful EBSD examination in the TMAZ is used to understand how the original grain structure transform to the SZ microstructure during the FSW process [11,12]. Thus, a grain boundary map was taken from TMAZ indicated by the black rectangle in Fig. 1(a), as shown in Fig. 3(a). Based on the careful EBSD observations, the following two important characteristics were identified: (1) Some HAB segments can be found in the grain interior, indicated by the black circle in Fig. 3(a). These HAB probably formed by the evolution of LAB in the vicinity of the initial grain boundaries. As the strain increased, the LAB gradually transformed into HAB and penetrated the original grain, ultimately creating fine grains. This

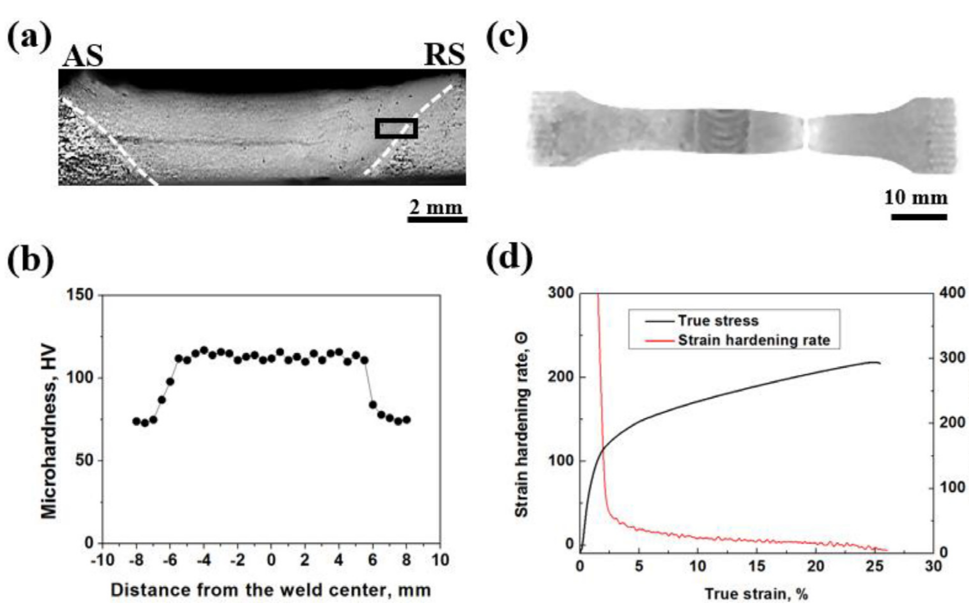

Fig. 1. (a) cross-sectional overview and (b) fracture location of the welded joint, (c) hardness profile measured across the weld mid-thickness, (d) true stress–strain curve and strain hardening rate vs. true stain of the LL-FSW AZ31B Mg alloy. "AS" and "RS" in (b) mean advancing side and retreating side of the weld, respectively.