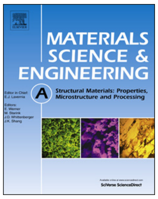
Materials Science & Engineering A 735 (2018) 336–342

Contents lists available at ScienceDirect

Materials Science & Engineering A

journal homepage: www.elsevier.com/locate/msea



The origin of different microstructural and strengthening mechanisms of copper and brass in their dissimilar friction stir welded joint

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ARTICLE INFO

Keywords:
FSW
Copper
Brass
EBSD
TEM
Nanoindentation

ABSTRACT

Pure copper and brass alloy with 30%wt Zn were successfully dissimilar joined using friction stir welding (FSW). The microstructure and mechanical properties of the copper and brass in the stir zone of the joint were characterized using light microscopy, electron backscattered diffraction (EBSD), transmission electron microscopy (TEM), and nanoindentation test. The results showed that both continuous and discontinuous dynamic recrystallizations (CDRX and DDRX) occurred during FSW. Higher values of DDRX in the brass caused (001)[110] shear texture with an intensity of 3.71, average grain size of 2.7 µm, 80% of high angle grain boundaries, Taylor factor of 3.01, and dense dislocation with tangle structures. On the other hand, $(1\bar{1}2)[110]$ and $(\bar{1}1\bar{2})[\bar{1}\bar{1}0]$ shear texture components with an intensity of 6.03, average grain size of 3.6 µm, 76% of high angle grain boundaries, Taylor factor of 2.99, and low density of dislocations with cell structures were formed in the copper due to the larger proportion of the CDRX. The lower n and higher σ_y values in brass (0.26/172.1 MPa) compared to those of the copper (0.37/76.2 MPa) were due to the higher amounts of strengthening mechanisms of grain boundaries and dislocations.

1. Introduction

Copper and its alloys have attracted considerable attention in both the academic and industrial areas due to their unique properties such as good combination of strength and ductility, high corrosion resistance, high thermal and electrical conductivities [1]. Therefore, the demands for welding and joining of the copper alloys have been increased. Unfortunately, the conventional fusion welding processes are not suitable for joining the copper alloys, which is due to two major reasons. First, the high thermal conductivity of the copper alloys causes the need for high heat inputs during fusion welding processes, and hence wide heat affected zone (HAZ) can be formed in the structure of the joints. Second, fusion and solidification of the weld metal results in dendritic structures, macro and micro segregations, porosities, inclusions, shrinkages, large distortions, residual stresses, etc. in the joints [2]. These difficulties of the conventional fusion welding have encouraged the scientists to find new methods for joining the copper alloys.

Fortunately, it has been proved by researchers that the friction stir welding (FSW) is a suitable method to join the copper and its alloys [3,4]. In this process, a rotating non-consumable tool inserts into the work pieces and then traverses along the welding line. During FSW, the materials do not melt, in which the coexistence of heat and severe plastic deformation leads to formation of sound joints. Thus, the conventional fusion welding problems, arisen from the melt and solidification steps, are eliminated by FSW. On the other hand, the FSW joints have a deformed and recrystallized structure, which typically causes enhancement of the mechanical properties [5,6]. Thus, FSW can be a promising method to replace the conventional fusion methods for joining the copper alloys.

Some researchers have studied the FSW of pure copper in recent years [7–17]. For example, Xu et al. [7] have used large axial load in conjunction with very low traverse and rotational speeds to join 2 mm thick pure copper plates by FSW. Their results showed that by using the mentioned condition, the strength of the joints could be increased without any reduction of elongation due to formation of fine grains with low dislocation density. Xie et al. [11] have used low heat input conditions, i.e. rotational speed of 400–800 rpm and traverse speed of 50 mm/min for joining 5 mm thick copper plates by FSW. They revealed that by using a rotational speed of 400 mm/min, the grain size of the stir zone (SZ) reduced to 3.5 µm. Sun et al. [16] have developed the processing window for FSW of 2 mm thick commercially pure copper plates. Their results showed that by FSW at a rotational speed of

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https://doi.org/10.1016/j.msea.2018.08.068
Received 2 May 2018; Received in revised form 18 August 2018; Accepted 21 August 2018

400–1200 rpm, traverse speeds of 200–800 mm/min, and axial loads of 1000–1500 kg, the defect free joints of pure copper could be achieved.

In addition to the copper, the FSW of copper base alloys, especially the brasses (Cu-Zn alloys) has attracted many attentions in recent years [18–26]. Wang et al. [19] have used water flowing during FSW of Cu-30%wt Zn, which caused the formation of grains inside the SZ of the joints with an average size smaller than 1 µm. They showed that by using water cooling during FSW, the mechanical properties of the joints could be improved considerably. Mironov et al. [21] have investigated the grain structure formation during FSW of Cu- 30%wt Zn alloy. They demonstrated that the new grains form by bulging of the grain boundaries and nucleation mechanism during FSW, which causes finer grain sizes and high strengths in the joints. Park et al. [25] have characterized the microstructural and mechanical properties of the friction stir welded Cu- 40%wt Zn plates. Their results showed that using lower heat input condition resulted in narrow HAZ, finer grain sizes, and higher hardness.

According to the above literature, some researchers have investigated FSW of similar pure copper or similar brass plates. However, an investigation into the dissimilar FSW of copper and brass plates is lacking. Therefore, the aim of this study was characterization of the microstructural and mechanical properties of the copper and 70/30 brass in their dissimilar friction stir welded joint. For this purpose, electron backscattered diffraction (EBSD), transmission electron microscopy (TEM), and nanoindentation test were employed.

2. Experimental procedure

Pure copper and brass (Cu- 30%wt Zn) plates with dimensions of 100 mm × 50 mm × 2 mm were used as the base materials (BMs). A H13 hot work steel tool consisted of a shoulder with a diameter of 12 mm, and a pin with a diameter and length of respectively 3 mm and 1.75 mm was employed. The copper and brass plates were placed in the advancing and retreating sides (AS and RS), correspondingly. The plates were friction stir welded at tool rotational and traverse speeds of 400 rpm and 100 mm/min, respectively. It is notable that this low heat input FSW condition was selected to obtain finer grain sizes according to the literature.

The microstructure of the joints was first examined by using a light microscopy (LM). The LM specimens were cross sectioned from the joints perpendicular to the FSW direction, and they were then prepared by mechanical polishing and etching with a solution of 50 mL HCl, 10 mL H_2O and 5 g $FeCl_3$. A Philips XL30 E-SEM field emission gun scanning electron microscope equipped with electron backscattered diffraction (EBSD) system was employed for OIM. The specimens for OIM were finalized after mechanical polishing by electropolishing for 30 s in a solution containing 250 mL H_3PO_4, 250 mL ethanol, 50 mL propanol, 500 mL distilled water, and 3 g urea at 10 V and 25 °C. The step size of 0.5 µm was used in EBSD scans. TEM (JEOL JEM 2010) was used for more clarification of the microstructural details. For TEM sample preparation, the electrojet thinning was used by means of a solution including 30% H_3PO_4 and 70% distilled water at the applied potential of 80 V. Moreover, for investigating the mechanical properties of the copper and brass in the SZ of the joint, nanoindentation tests were performed using MTS XP Nano-indenter equipped with a Berkovich diamond indenter. Minimum 20 indentations were conducted for each weld zones at the maximum load of 40 mN. The reverse algorithm developed by Dao et al. [27] was used to extract the local mechanical properties (yielding strength σ_y and strain hardening exponent n) from the nanoindentation data.

3. Results

3.1. Microstructural zones

The cross section macrostructure of the joint is shown in Fig. 1a. As seen from Fig. 1a, it had different microstructural zones of BM, thermomechanically affected zone (TMAZ), and SZ in both sides of the copper and brass. It is notable that HAZ was not seen may be due to the high thermal conductivity of the copper and brass or because of the low heat input condition during FSW (i.e. 400 rpm and 100 mm/min). The BMs of copper and brass had coarse equiaxed grains, which was due to the initial annealing process of them (Fig. 1b and c). In addition, the TMAZs had similar structures of deformed grains (Fig. 1d and e). In TMAZ, the grains have been elongated along the shear stresses induced by the rotational tool. In addition, due to the insufficient heat and deformation in TMAZs, the dynamic recrystallization (DRX) has not been occurred, and hence the grains are just deformed in these areas. On the other hand, the SZs of copper and brass have contained fine equiaxed grains, which shows the occurrence of DRX in these areas [17,18]. The detailed microstructures and mechanical properties of the copper and brass in the SZ of their dissimilar joint will be discussed in the following sections.

3.2. OIM

The EBSD data of the copper and brass in the SZ of the joint (indicated in the Fig. 1a) were evaluated by TSL OIM software. The inverse pole figure (IPF) map, grain boundary (GB) map, in conjunction with grain boundary characterization distribution (GBCD) of the copper and brass are illustrated in Fig. 2a-d. The copper had an average grain size (area fraction) of 3.6 µm, where that of the brass side was 2.7 µm. The amount of high-angle grain boundaries (HAGBs), low-angle grain boundaries (LAGBs), and twin boundaries (TBs) were 51–24–25% and 55–20–25% correspondingly for copper and brass SZs. The grain average misorientation (GAM) and Taylor factor maps of the copper and brass in the SZ are shown in Fig. 3. When the GAM values are larger, it means that the structure is more deformed, and hence the dislocation density in more [28]. In Fig. 3a and b, the different GAM values from 0° to 2° are represented by a blue gradient color. The average Taylor factor was calculated from EBSD data (Fig. 3c and d) for the both copper and brass in the SZ as 2.99 and 3.01, respectively. The Taylor factor, which is influenced by the texture of the material, affects the mechanical properties.

3.3. Texture

For the texture analyzes of the SZ, the acquired pole figures (PFs) were rotated along the rolling direction (RD), traverse direction (TD), and normal direction (ND) according to the method developed by Fonda et al. [29]. The final rotated PFs and corresponding orientation distribution functions (ODFs) for the copper and brass in the SZ have been illustrated in Fig. 4. It is well documented that FSW causes formation of the shear texture components in the metals and alloys. Comparison between the texture results (Fig. 4a and b) and simple shear texture components of the face centered cubic (FCC) metals (Fig. 4c) shows that B_1 i.e. $(1\bar{1}2)[110]$ and B_2 i.e. $(\bar{1}1\bar{2})[\bar{1}\bar{1}0]$ in copper were the shear texture components. However, in the case of the brass, the C i.e. (001)[110] was the main shear texture components. Moreover, the texture intensity for the copper and brass were 6.03 and 3.71, respectively.

3.4. TEM

The TEM images of the copper and brass in the SZ are illustrated in Fig. 5. Two important results can be obtained from Fig. 5. First, the dislocation densities were different in copper and brass. In brass, the dislocation density was higher than that of the copper in SZ. Second, the dislocation structure in the case of copper was cell structure, where in the case of brass, they had the tangle structures. It can be concluded that the recovery of the dislocations has been occurred in the copper due to its higher stacking fault energy (SFE). However, the low SFE of