Microstructural Variation and Tensile Properties of a Cast 5083 Aluminum Plate via Friction Stir Processing

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A significant friction stir processed defect of kissing bond accompanied with coarse abnormal growth grains can be recognized where located aside the bottom of advancing side as raised up the rotating speed higher than 1450 rpm. Friction stir processed specimens not only possess refined α-Al phase, the coarse irregular solidification precipitates as breakup Al(Mn,Fe) and Al6Mn actually could be refined and homogenized in stir zone result in a more stable strain hardening behavior especially from the initial stage of tensile deformation, abovementioned factors significantly benefit the tensile ductility of friction stir processed cast plate, but microstructural refinement did not cause apparent difference in yield strength and hardness performance that might result from the textural feature of friction stir processed fully annealed 5083 casting material. [doi:10.2320/matertrans.M2009049]

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1. Introduction

High strength Al-Mg alloys, such as alloy 5083, have been widely used especially for the vacuum chamber or frame structure of vapor deposition substrate where stable mechanical properties at elevated temperature are required. For example, fully annealed Al-Mg alloy sliced thick plate, which is cut from large size casting slab with comparatively isotropic and equaxed-grained structure, mainly has been applied to the frame material inside the heating chamber as a carrier in TFT-LCD industry. In practical applications, large and slowly cooled 5083 casting slab possesses coarse solidification microstructure especially characterized by coarse Mn-rich intermetallic compounds is pertaining to induce poor surface quality or embrittlement under subsequent forming process. Furthermore, this class of casting alloy is difficult to join by conventional fusion welding because it will seriously weaken the mechanical properties. In order to overcome abovementioned problems, previous reports1-3) have suggested that grain refinement can raise formability. The main aim of this investigation was to determine if friction stir process can be applied to the specific location of large size frame type components where to enhance the mechanical properties is required.

Considering the friction stir welding (FSW)4-11) have emerged as a promising solid-state joining process. Particularly, when FSW utilized in tailor weld blanks, a post-weld deformation such as pressing and bending works is inevitable. Friction stir processing (FSP)12-14) which is essentially similar to FSW, can be applied to surface modification or a specific local areas of components where better mechanical properties are required. As it has been developed based on basic concept of FSW, this process is often utilized to refine the grain size or to homogenize the heterogeneous intermetallic solidification precipitates. Many previous reports5,6,14) characterizing the microstructural feature pertaining to dynamic recrystallization have been completed.

Table 1 Chemical compositions of AMC (mass%).

<table>
<thead>
<tr>
<th></th>
<th>Mg</th>
<th>Mn</th>
<th>Cr</th>
<th>Fe</th>
<th>Si</th>
<th>Zn</th>
<th>Cu</th>
<th>Ti</th>
<th>V</th>
<th>Al</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>4.86</td>
<td>0.58</td>
<td>0.06</td>
<td>0.27</td>
<td>0.07</td>
<td>0.04</td>
<td>0.04</td>
<td>0.01</td>
<td>0.01</td>
<td>bal</td>
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</table>

An as received Al-4.8Mg casting (AMC) slab (950 × 950 × 1200 mm) undergoing a hot top solidification process was used in this study after sliced to 10 mm thickness plate, and its chemical compositions were listed in Table 1. The dimension of tool was 15 mm in shoulder diameter, 8 mm in pin diameter and 3 mm in pin length. The parameters of FSP were 1050 to 1650 rpm in tool rotating speed, 0.3 mms⁻¹ in processing speed, and 1° in tool tilted angle.

Dogbone tensile specimens were machined along the tool processed direction (PD) with 10 mm in gage length and 2.2 mm in width. Note that PD, ND, and TD denote the processed, normal, and transverse direction of the plate. Tensile tests were performed at room temperature and the initial strain rate was 1.67 × 10⁻³ s⁻¹ in a constant crosshead speed. Dynamic strain aging (DSA)15-18) phenomenon and strain hardening behavior19-21) were examined from the obtained stress-strain curve, the feature of serrated flow was characterized by the plastic onset strain of serrations and the magnitude of stress drop. For strain hardening behavior, strain hardening rate δs/δεp20,21) where σs and εp are true stress and true plastic strain, was investigated in the period of uniform plastic deformation.

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Textural variation due to FSP was examined by Rigaku D/MAX2500 X-ray diffractometer. The SZ sample was directly got from the untested tensile specimen gage at the dimension of 2 mm × 2 mm and AMC was got from the base metal at dimension of 10 mm × 10 mm. The different sizes of XRD specimens were chosen for the reason to make the amounts of detected grains between SZ and AMC as similar as possible. In addition, the orientation distribution functions (ODFs) were calculated from the incomplete pole figures \{111\}, \{200\} and \{220\}. The ODFs were carried out by using Bunge’s series method, and represented in Euler’s space at constant \( \Phi_2 \) sections. The textural components are occasionally labeled as (hkl)[uvw], where (hkl) is the crystallographic plane perpendicular to ND, and [uvw] is the crystallographic direction parallel to PD. The textural variation due to FSW/FSP can be referred to many previous reports of Sato et al.,22–24) but the effects on the tensile properties are still unclear. In this paper the average Taylor factor\(^{25}) was calculated to understand that.

3. Results

3.1 Microstructural feature

Figure 1 exhibits fully annealed casting microstructure of large size casting slab, characterized by solidificational equaxed grains, significantly coarse inhomogeneous-distributed \( \text{Al}_6 \text{(Mn,Fe)} \) on the solidification cell boundary and some needle-like \( \text{Al}_6 \text{Mn} \) precipitating inside the grains. An one-pass-FSPed (1050 rpm in rotating speed) specimen possesses a microstructural feature that characterized by recrystallized fine grains and uniformly distributed spherical particles as \( \text{Al}_6 \text{(Mn,Fe)} \) and \( \text{Al}_6 \text{Mn} \) due to an intense breakup of the intermetallic compounds as shown in Fig. 2. From Fig. 2, friction stir processed area revealed two different zones (i.e. SZ1 and SZ2), it should be noted that a number of material flow are observed within SZ2 area as shown in Fig. 3, however, the coarse equaxed grains of AMC can be effectively refined from the order of 80 \( \mu \text{m} \) to about 10 \( \mu \text{m} \) (Fig. 1(a), Fig. 2(e) and 2(f)). From a typical evidence as shown in Fig. 2(d), the breakup \( \text{Al}_6 \text{(Mn,Fe)} \) particles actually can be homogenously stirred into the grains and the grain boundary, while the microstructural feature of as cast sample (AMC) consists of coarse interdendritic irregular \( \text{Al}_6 \text{(Mn,Fe)} \) particles as shown in Fig. 1(b).

On the other hand, the FSPed samples have susceptibility to kissing bond defect accompanying with abnormal growth grain whereon the advancing side as raising the rotating speed up to 1450–1650 rpm (Fig. 3). The macrostructure, however, reveals significant difference on material flows which are generally inhomogeneous as shown in SZ2 area and it still presents up to 1650 rpm. This result indicates that the microstructural feature of friction stir processed zone is splitted into a bimodal distribution of finer grain band and coarser grain as shown in Fig. 3. This may result from a relatively high deformation temperature during friction stir process, especially may occur as the material flow relating to kissing bond which locates in the vicinity of interface of base material. The increase of deformation temperature with increasing rotation speed (1450–1650 rpm) is compatible with the observation that revealed abnormal growth coarse grain and kissing bond defect as shown in Fig. 3(g) and 3(h). No other cracking was observed in any of the locations exposed in the vicinity of friction stir processed area.

3.2 Microhardness profile and tensile properties of friction stir zone

From microhardness profile as shown in Fig. 4, the microhardness of friction stir processed specimen are quite insensitive to microstructural variation and elevated temperature exposure pertaining to frictional heat input. Figure 5 showed the comparison of tensile properties of AMC, SZ1 and SZ2 respectively, it can be verified that the presence and different morphology of second phase and grain size, it should be noted that the SZ1 (15 \( \mu \text{m} \)), SZ2 (12 \( \mu \text{m} \)), and AMC (80 \( \mu \text{m} \)) samples also performed very little difference on yield stress, particularly did not exhibit appreciable dependence on grain size. On the other hand, comparison of SZ specimen and AMC specimen indicates that both of the average uniform elongation and total elongation is quite different. For example, the total elongation of base metal is much lower than that of the friction stir processed specimen (SZ1 and SZ2). The initial intermetallic compound size prior to FSP was larger than 25 \( \mu \text{m} \) and the initial grain size was partially larger than 100 \( \mu \text{m} \) as shown in Fig. 1, the grain size or particle size, however, obvious larger than that of friction stir processed specimen.
Figure 6 exhibited the feature of stress-engineering strain curves of each specimen, very similar shapes of stress-strain curves were obtained from the repeated tensile testing, a serrated flow caused by dynamic strain aging were recognized. From Fig. 6, the critical onset strain of SZ1 and SZ2 specimens both are smaller than AMC as the indication of the arrows. On the other hand, comparison of the stress amplitude of serrations indicated that the SZ1 specimens performed larger stress drop than AMC. Furthermore, from Fig. 7, similar trends of decreasing the strain hardening rate with true plastic strain are observed within uniform plastic deformation region respectively. It is evident that the strain hardening rate of AMC is significantly enhanced only in the initial deformation stage, however, show lower tensile elongation than SZ specimen. This may be attributed to the grain size and the morphology difference of IMC phase.

Based on Fig. 5, though the yield stress is reduced after a single pass FSP, the large tensile ductility in the specimen after FSP can be attributed to significant strain hardening after yielding. In contrast, the AMC specimen shows that lower strain hardening capacity is associated with mechanical instability due to the presence of coarse Al<sub>6</sub>(Mn,Fe) and Al<sub>6</sub>Mn compounds resulting in a low tensile elongation.

4. Discussion

The microstructural refinement and homogeneity obtained by friction stir process are always the strengthening factors for yield strength and hardness. Our previous studies have reported that the effect of grain size on yield stress of the as-stirred 5052 specimen manufactured by different rotation speed (500–1500 rpm), however the experimental results actually follow the Hall-Petch equation ($\sigma_y = 176d^{-1/2} + 64$) and are linearly proportional to the grain size. As compared to fully annealed 5083 cast slab used in this study, which is also a kind of non-heat-treatable aluminum alloy with
inhomogeneously large grains, though grain size refinement can be acquired by FSP, a negative effect presents as shown in Fig. 5 which would result in the microstructural feature of FSP while textural effect plays an important role on tensile yield stress. Figure 8 exhibits the comparison of textural feature between AMC and SZ specimens, in which obvious
Fig. 6 Tensile engineering stress-flow strain curve.

Fig. 7 Strain hardening rate-true plastic strain diagram (in the period of uniform deformation).

Fig. 8 Comparison of textural feature between AMC and SZ specimens: (a) ODFs of AMC, (b) ODFs of SZ, (c) IPF of AMC’s PD plane, (d) IPF of SZ’s PD plane, (e) Contours of constant Taylor factor M for axisymmetric flow based on [111](110) slip.25
difference can be recognized. From the ODFs (Fig. 8(a) and 8(b)), the textural components of AMC and SZ can be defined and listed with the individual Taylor factor in Table 2. In addition, the IPFs (inverse pole figures) of PD plane are shown in Fig. 8(c) and 8(d). Since comparing IPFs to the contours of constant Taylor factor, M (Fig. 8(e)) could not clearly identify the difference between AMC and SZ, the fractions of major textural components were approximated by the intensity level-weighted method to calculate the average Taylor factor, \( \bar{M} \). The textural intensity levels are represented as the contours in ODFs which decreases gradually from the center and the second, third, etc. contours which are contributed from the textures similar to center are also taken to consider in this paper. Taking AMC for example, \((112)[021]\) is the primary texture with 2636.7 in highest level and 2264.3, 1891.8, 774.4, 402.0 in following contours, the five levels are summed as the weight for the fraction of \((112)[021]\), however, other levels are so weak to be ignored. In the same way, summation of 774 and 402 is used as the weight for the fraction of the secondary texture \((4, 15, 8)[943]\). By the abovemented method, the approximated textural fraction can be defined and used for \( \bar{M} \) value calculation. The \( \bar{M} \) values of AMC and SZ for axisymmetric flow based on \([111][100]\) slip are 2.95 and 2.89. The lower \( \bar{M} \) value means the textural variation is a disadvantageous factor for tensile yield stress of friction-stirred cast 5083 aluminum alloy.

It is evident from Fig. 5 that tensile ductility is significantly enhanced after the single pass FSP. The tensile elongation data of the AMC is notable lower (about 10%) due to the presence of coarse primary \( \alpha \)-aluminum equaxed grains, irregular shape \( \text{Al}_6(Mn,Fe) \) compounds and solidification defects, this is in good agreement with many previous reports of cast alloys.\(^9,26\) Based on abovementioned experimental results, it is suitable to propose that the ductility improvement can be acquired by friction stir process due to the refinement and homogeneity of coarse casting structure.

Comparison of Fig. 6 indicated that the tensile deformation behavior of AMC and SZ specimens also can be examined from the difference of tensile flow curve, based on the characteristics of serrations, smaller critical onset strain and larger stress amplitude imply that the DSA effect were promoted during tensile deformation. Due to the finer dynamic recrystallization grain boundary and friction stirred breakup \( \text{Al}_6(Mn,Fe) \) particle will provide larger retardation on the movement of dislocations, however, the beneficial condition to satisfy critical condition of ambient atmosphere of dissolved atoms. From previous reports,\(^27\) friction stir processed specimen possesses higher retention of dislocation density and finer subgrain size than fully annealed metal, both factors are in agreement with the well-known promotive factor for the occurrence of serrations.

On the other hand, while plastic deformation undergoing, higher and more stable strain hardening behavior reduced the local deformation and it is beneficial to acquire larger uniform elongation (UE). AMC initially performed higher strain hardening rate than SZ1 and SZ2, but suddenly dropped to the lowest (Fig. 7). It meant that SZ1 and SZ2 possessed much more stable strain hardening behavior than AMC. The difference of tensile strain hardening rate between the specimens without and after FSP can be recognized that resulted from the remarkable difference of IMC morphology as shown in Fig. 1 and Fig. 2, and this result also in good agreement with previous reports of Liu et al.\(^20\) and Poole et al.\(^23\) They proposed that the drastic drop of strain hardening rate was relative to the stress concentration of particles with high aspect ratio in the Al matrix composite. The microstructure of SZ specimen, however, becomes reasonably homogeneous after FSP, spherization of \( \text{Al}_6(Mn,Fe) \) and \( \text{Al}_6Mn \) (Fig. 2(a)–2(c)) generated by FSP introduced the load transfer inefficiency and therefore did not generate enough stress to break the particles or to debond the interface, so the strain hardening rate of SZ1 and SZ2 demonstrate a more stable strain hardening behavior than AMC.

In addition, more significant occurrence of serrations resulting from microstructure feature of FSP is also one of promotive factor for strain hardening behavior during all the uniform stage of tensile deformation, the subsequent necking phenomenon can be suppressed that pertaining to the performance of tensile ductility behavior.

Non-uniform elongation (NUE), which was the value of the subtraction from TE with UE (Fig. 5), denoted the ability of continued deformation after necking arising in tensile test. NUE of AMC, SZ1 and SZ2 was 1.7%, 7.1%, and 8.3%. In order to find the reason, the subsurface of tensile specimens near the breakage was observed. Figure 9 showed the fracture (indicated with the arrows) on \( \text{Al}_6(Mn,Fe) \) in all of the broken tensile specimens, and it meant that \( \text{Al}_6(Mn,Fe) \) fracture was the starting point leading to the final break. In AMC the initial fracture generated easily and grew rapidly on the large \( \text{Al}_6(Mn,Fe) \) particles; this manner by contrast with spheroidized finer \( \text{Al}_6(Mn,Fe) \) in both specimens of SZ1 and SZ2 that showing very little fractured evidence, so it could sustain more deformation (i.e. higher NUE). Furthermore, comparing Fig. 9(b) and 9(c) found that the \( \text{Al}_6(Mn,Fe) \)

<table>
<thead>
<tr>
<th>AMC</th>
<th>( \text{Al}_6(Mn,Fe) )</th>
<th>( \text{Al}_6Mn )</th>
<th>( \text{Al}_6Mn )</th>
<th>( \text{Al}_6Mn )</th>
<th>( \text{Al}_6Mn )</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>(112)[021]</td>
<td>402.0–2636.7</td>
<td>2.94</td>
<td>A</td>
<td>(110)[001]</td>
</tr>
<tr>
<td>B</td>
<td>(4, 15, 8)[943]</td>
<td>402.0–774.4</td>
<td>2.97</td>
<td>B</td>
<td>(12, 9, 4)[340]</td>
</tr>
<tr>
<td>C</td>
<td>(123)[274]</td>
<td>19.9–29.5</td>
<td>2.98</td>
<td></td>
<td></td>
</tr>
<tr>
<td>D</td>
<td>(132)[111]</td>
<td>19.9</td>
<td>3.67</td>
<td></td>
<td></td>
</tr>
<tr>
<td>E</td>
<td>(011)[100]</td>
<td>19.9</td>
<td>3.67</td>
<td></td>
<td></td>
</tr>
<tr>
<td>F</td>
<td>(225)[322]</td>
<td>19.9</td>
<td>3.46</td>
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</table>
particles in SZ2 was more dispersive than SZ1, it meant larger mean free path for particle adjoining to cause final break, hence, the NUE of SZ2 was larger than SZ1.

5. Conclusion

1. The kissing bond defect can be recognized within stir processed zone on the advancing side as increased the rotating speed from 1250 rpm. An abnormal grain growth accompany with the occurrence of kissing bond crack can be observed. However, friction stir processed 5083 commonly revealed two different zones and material flow feature that pertains to the difference of frictional heat input. The breakup coarse Al₆(Mn,Fe) and Al₆Mn particles caused by slow solidification rate actually could be homogenously dispersed and stir into the grains or grain boundary by FSP.

2. Serrations occurred during tensile deformation, the density of serrations and the amplitude of stress drop can be correlated with the strain hardening behavior. However, friction stir processed specimen indicated more steady strain hardening behavior during the tensile deformation, which will be of much benefit to improve tensile ductility.

3. Little difference of tensile yield stress between AMC and SZ specimens despite significant microstructural refinement that pertains to the textural feature of friction stir processed zone.

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REFERENCES