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Comparative study on local and global mechanical properties of bobbin tool and conventional friction stir welded 7085-T7452 aluminum thick plate

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ABSTRACT: 7085-T7452 plates with a thickness of 12 mm were welded by conventional single side and bobbin tool friction stir welding (SS-FSW and BB-FSW, respectively) at different welding parameters. The temperature distribution, microstructure evolution and mechanical properties of joints along the thickness direction were investigated in present work, and digital image correlation (DIC) was utilized to evaluate quantitatively the deformation of different zones during tensile tests. The results indicated that heat-affected zone (HAZ), the local softening region, was responsible for the early plastic deformation and also the fracture location for SS-FSW samples, while a rapid fracture was observed in weld nugget zone (WNZ) before yield behavior for all BB-FSW specimens. The ultimate tensile strength (UTS) of SS-FSW joints presented the highest value of 410 MPa, as 82% of the base material, at a rotational speed of 300 rpm and welding speed of 60 mm/min, much higher than that of BB-FSW, the joint efficiency of which only reached 47%. This should be attributed to the Lazy S defect produced by a larger extent of heat input

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during the BB-FSW process. The whole joint exhibited a much higher elongation than the slices. SEM analysis of the fracture morphologies showed that joints failed through ductile fracture for SS-FSW and brittle fracture for BB-FSW.

Keywords: Aluminum alloy; Friction stir welding; Bobbin tool; Temperature distribution; Microstructure; Mechanical properties

1. Introduction

Bobbin tool friction stir welding (BB-FSW) is a type of solid-state welding technology based on the principle of conventional FSW [1-4]. During the BB-FSWing process, the bobbin tool consists of two shoulders connected by the tool pin, the lower shoulder replaces the backing plate used in conventional FSW [1], which allows for welding machines with substantially lower stiffness and adds extra flexibility of FSW. It is possible to carry out the FSW of closed profiles and complex shaped structures by using bobbin tool. As the bottom shoulder also functions as heat source, the temperature gradient is significantly reduced along the thickness direction. In addition, the ability to weld a workpiece simultaneously from both sides makes BB-FSW an effective technology to eliminate the occurrence of root flaw [5,6].

For the FSW, the temperature field determines not only the flow behavior of base material (BM) but also the microstructure evolution of different weld regions, and further influences the mechanical properties of joints [3,7]. Until now, investigations on the temperature distribution, microstructure and mechanical properties of BB-FSWed joints have focused mainly on 2000 [6,8,9] and 6000 [10-13] series aluminum alloys. A thermo-mechanical model of the BB-FSW for 2014 aluminum alloy was established by Liu et al [14], and simulation results indicated that the temperature field of the weld cross section presented symmetry approximately about the mid-thickness of the work piece, and the peak temperature for the point at retreating side (RS) was higher than advancing side (AS). Wan et al. [12] found that the typical weld shape of the BB-FSW joint differed from the conventional FSW joint, and was roughly hourglass-shaped. Zhang et al. [6] carried out BB-FSW of 2A14 aluminum alloy, and the weld formation and mechanical properties of the joints were investigated, achieving a maximum strength efficiency of 75%. On the whole, there have been relatively few studies on the development of BB-FSW of high-strength aluminum alloy plate especially on the inhomogeneity of microstructure and joint properties along the thickness direction.

Al-Zn-Mg-Cu alloy with a brilliant application prospect has been already widely used in aerospace fields to achieve lightweight [15,16] due to its low density, high strength and so forth. Therefore, in this paper, 7085 aluminum alloy plates were welded by SS-FSW and BB-FSW, respectively, and the focus is placed on the temperature distribution, weld formation, microstructure and mechanical properties along the thickness direction of the two types of joints.

2. Material and experimental procedures

The BM used in the present study was a 12 mm thick 7085-T7452 aluminum alloy plate. The nominal chemical compositions of the BM is listed in Table 1. The plates were butt welded along the longitudinal direction of the samples by a medium-sized gantry FSW machine (FSW-LM-E154). For the SS-FSW, the welding tool is made of die steel H13 with the shoulder of 24 mm in diameter, and a conical threaded pin of 11.8 mm in length, with 10.5 mm and 6 mm in root and tip diameter, respectively. The bobbin tool has scrolled feature on both shoulders and three flats feature on the cylindrical pin with thread. The two shoulders are made of die steel H13 and have a diameter of 34 mm, while the pin is made of high-temperature alloy MP159 with the diameter of 13 mm and length of 11.8 mm, as shown in Fig. 1. Based on previous studies, the welding parameters selected to produce various joints are shown in Table 2.

The temperature evolution during the welding process was detected by K-type thermocouples plugged in feature points at different positions of the plates. In order to protect thermocouples from destruction by the tool, blind holes of 1.5 mm in diameter and 25 mm in depth were designed. Fig. 2 shows the positions of blind holes, which allow for the measurement of the temperature changing process with the AS (by points 1-9) and RS (by points 10-12) of welds, the outside of pins (by points 7-10) and the thickness direction of work pieces (by points 1-9).

After welding, the specimens for metallographic analysis were cut perpendicular to the welding direction, and then grinded with #240-#7000 sandpaper, polished using a diamond paste and etched with Keller's reagent (3 mL HNO_3, 6 mL HCL, 6 mL HF and 150 mL H_2O). Macro morphology was observed using a laser scanning microscope (OLS4000), and microstructure analysis was performed by an optical microscope (OLYMPUS GX71) and the Image-Pro Plus software to measure the grains size. In this paper, the cross section of weld is divided into three slices along the thickness direction, which are top layer, middle layer and bottom layer, respectively, to draw a comparison analysis of the microstructure evolution for two types of FSW welds.

The 2D microhardness maps were obtained at the whole weld region on the polished cross sections using a Vickers hardness tester (THV-1D) with a testing load of 0.2 kg and a dwell time of 20 s. The spacing between adjacent indentations was 0.5 mm. The tensile samples were prepared according to ASTM E8 with a gauge of 32 mm long by 10 mm wide, and the specimens were cut into equal three slices. Tensile properties of both the whole and slices of joints were evaluated with two tensile specimens cut from the same joint. The room temperature tensile tests were carried out at a strain rate of 10^{-3} s^{-1} using an electron universal testing machine (Instron-3382), and the local strain fields were determined by digital image correlation (DIC). Each specimen in this test was prepared by applying a random speckle pattern on the cross section using black and white spray paints. During the test, the specimen was photographed every half second by CCD cameras. After tensile test, the fracture features of the specimens were observed by a scanning electron microscopy (X-mzx20/INCA 250).

3. Results and discussion

3.1 Thermal analysis

The transient temperature curves of SS-FSW and BB-FSW processes are presented in Fig. 3. On the whole, the variation tendencies of curves for both welding processes were essentially consistent, the initial heating rate and peak temperature of BB-FSW, however, were higher than those of SS-FSW, which agrees well with previously reported results [9,13]. Fig. 4 shows the peak temperatures of different thermocouple points during the welding process, it can be seen that the highest peak temperature with a value of ~510 ℃ was recorded at point 9 in the AS of the BB-FSW joint under the rotational of 200 rpm and welding speed of 150 mm/min. For the SS-FSW joints, the corresponding value was much lower (~455 ℃) under the rotational of 300 rpm and welding speed of 60 mm/min, respectively. These facts indicate that bobbin tool provides more heat input during the welding process due to its lower shoulder, the stirred zone, therefore, experienced a higher thermal cycle and more severe plastic deformation during the BB-FSW process.

According to Fig. 4, it can be observed that the peak temperature in the AS of the SS-FSW joints decreased gradually with increment of the distance from blind holes to the top surface and the weld line, respectively. The peak temperature in the RS was lower than that of the same position in the AS. The corresponding values measured at points 12, 11 and 10, for instance, were lower by 31 ℃, 29 ℃ and 16 ℃ than those of points 2, 5 and 8, respectively, at the rotational speed of 300 rpm. It is attributed to the fact that the welding direction was the same to the linear velocity direction of rotating shoulder in the AS, and thusly the tool experienced higher frictional resistance and plastic shear force, leading to the material in the AS obtained more friction heat than that in the RS. The temperature distribution of BB-FSW joints along the thickness direction, by contrast, was of much difference. The peak temperature of lower shoulder affected zones was slightly higher than that of upper shoulder affected ones. With the rotational speed of 200 rpm, for example, the highest temperatures recorded at points 3, 6 and 9 were higher by 6 ℃, 4 ℃ and 5 ℃ than those of points 1, 4 and 7. As there is no backing plate used in BB-FSW, the main heat-dissipating form for the lower affected zone is the air convection [6], while the heat generated by upper shoulder can be dissipated by not only air convection but transferring into the spindle of welding machine, contributing to the temperature gradient along the thickness direction mentioned above. Moreover, the peak temperatures in the RS and AS of the corresponding positions followed the same law as SS-FSW joints, and it is noted that the thermal cycle of BB-FSW joints was more influenced by rotational speed.

3.2 Weld formation

Fig. 5 displays the cross section macrographs of SS-FSW and BB-FSW specimens produced by using two sets of welding parameters, respectively. For each SS-FSW weld, the AS is on the left side while the RS is on the right in the figure and through the whole paper, which differs from the BB-FSW welds due to opposite rotation directions of two tools. The SS-FSW weld exhibited a bowl-like shape, the top cross section of weld corresponded with an approximate width of the shoulder diameter and the weld bottom possessed the smallest width of the tip diameter of the pin. By contrast, the cross section of BB-FSW samples was nearly dumbbell shaped, characterized by wider upper and lower layers (comparable to the diameter of shoulders) and a smaller middle layer (greater than the diameter of the pin), which was attributed to the fact that the top and bottom surface of BM experienced the maximum deformation and frictional heat during BB-FSW process.

The macrostructure of the BB-FSW weld is more symmetrical compared with SS-FSW weld due to the use of a symmetrical bobbin tool. Moreover, the three distinguishing regions formed during both FSW processes, i.e. the WNZ, TMAZ and HAZ are indicated in Fig. 5. It is worthwhile to note that there is a rather sharp transition between the WNZ and the TMAZ in the AS of each weld, which is indicated by red arrow in Fig. 4, while a much smoother interface shows in the RS, which is consistent with phenomena recorded in previous studies [6,17,18]. This is because both flow behaviors and degrees of plastic deformation are different on the two sides of weld (AS and RS). To be more specific, the rotation of the tool was in the direction of welding in the AS, resulting in BM plastically deformed forward. The moving of weld material, however, was in the opposite direction under the effect of extrusion, and then larger relative deformation difference was developed between the BM and weld, accounting for that the interface in the AS was more obvious. While both flow directions of the undeformed and weld material were consistent in the RS, giving rise to a more diffuse transition region.

According to Shen et al. [19] and Zhang et al. [20] previous studies, the stir zone of welds was divided into the shoulder dominated zone (SDZ) and the pin-significant with shoulder-partial dominated zone (PSZ) as the Fig. 5b and 5c illustrate. The local material flow of the PSZ increased slightly for both SS-FSW and BB-FSW when increasing the rotational speed, leading to the dimension change of welds. Additionally, the appearance of S line through the WNZ is presented in Fig. 5d at the rotational and welding speeds of 200 rpm and 150 mm/min, respectively. It probably originated from the oxidized layer on the 7085 aluminum alloy surface, which was produced by higher thermal cycle at the rotational speed of 200 rpm, and the S line defect might give rise to the weakness of mechanical properties.

3.3 Microstructure characteristics

High magnification images of grain morphologies of BM and three different weld zones are presented in Fig. 6, which were obtained by the optical microscopy. Compared with the BM exhibiting slightly elongated grains, the WNZ affected directly by the tool showed small equiaxed recrystallized grains (Fig. 6b and 6c), since the material experienced a high degree of plastic deformation and frictional heating simultaneously. The TMAZ was located at edges of the pin, and thus equiaxed recrystallied grains mixed partially with rotated elongated grains (Fig. 6d and 6e) were produced due to the relative insufficient plastic deformation and the plastic shear stress of the flowing material during the welding operation. The HAZ, however, was only influenced by the thermal cycle but no stirring mechanically of the tool. Therefore the grain orientation was consistent with the BM while the grains were coarsened partially (Fig. 6c and 6g), resulting from the annealing treated microstructure of HAZ.

Fig. 7 shows the grains characteristics of the upper, middle and lower layers of WNZs along the thickness direction for both FSW techniques. It is observed that grains in the upper layer of WNZ of SS-FSW joint were larger than the ones in the lower layer (Fig. 7a-7c). The average grain size of upper layer, measured with the average linear intercept method, was 6.05 μm while the middle and lower regions exhibited an average grain size of 4.87 μm and 3.56 μm, respectively, at the rotational speed of 300 rpm. This is because the heat input from the single side tool decreases gradually through the thickness, leading to the grain size presents a decline trend from the upper to lower layer of WNZ. Unlike the SS-FSW specimen, the grains in the upper layer of WNZ of the BB-FSW samples were slightly smaller than those in the lower layer (Fig. 7d-7f), corresponding to the results of temperature distribution, indicating that the BB-FSW technique resulted in a more uniform grain structure in the WNZ along the thickness direction. For instance, when the rotational speed was 150 rpm, the upper, middle and lower layers had an average grain size of 6.34 μm, 6.22 μm and 6.53 μm, respectively. It can be concluded that the variations in grain evolution along the thickness were more apparent for the SS-FSW samples than those of BB-FSW.

Quantitative data on the grain sizes of WNZ through the thickness direction at different rotational speeds are presented in Fig. 8. For both FSW techniques, higher rotational speeds gave rise to a larger grain size, and specifically, the average sizes of recrystallized grains of WNZs for SS-FSW and BB-FSW increased from 4.76 μm and 6.07 μm to 5.71μm and 7.39 μm, with the increase of rotational speeds from 300 rpm and 150 rpm to 600 rpm and 200 rpm, respectively. It is related to a larger heat input provided by a higher rotational speed.

3.4 Mechanical properties

3.4.1 Microhardness distributions

The microhardness distribution of the joints is mainly associated with the evolving characteristics of grains and the state of precipitation phases during the welding process [21-23]. The maps of microhardness distributions in the whole cross sections for BM and all the joints obtained in the present work are presented in Fig. 9. It can be noted that the microhardness profiles exhibited a typical “W” shape along the cross section for both of SS-FSW and BB-FSW samples, the same as the phenomenon observed by previous studies in SS-FSW and BB-FSW aluminum alloys [1,6,7,24-27]. The significant softening region tended to be located at the HAZ as a result of the grains coarsening and the dissolution of precipitation phases. It is well known that the local softening region is usually consistent with the failure position during the tensile tests. Compared with microhardness values of BM scattered between 150 and 165 HV, the microhardness in stir zones of all the joints WNZ dropped significantly. The average microhardness of WNZ was further higher than that of TMAZ and HAZ, and it is attributed to the fine equiaxed microstructure in the WNZ produced by a significant effect combined of the thermal cycle and plastic deformation. Moreover, the microhardness values of HAZ in the AS were slightly lower than those in the RS, which was related to the finer grains in the RS. For SS-FSW joints, the minimum microhardness showed a decreasing trend when increasing the rotational speed, indicating an enhanced negative effect of the thermal cycle on mechanical properties of joints, while a reverse trend was observed for BB-FSW joints when the rotational speed was increased from 150 rpm to 200 rpm. It is also noteworthy that the heterogeneity of microhardness distributions was observed in the thickness direction for SS-FSW joints, which was commonly reported by other researchers [28,29]. In contrast, there existed a roughly symmetrical microhardness distribution of different layers through the thickness direction for BB-FSW joints. During the BB-FSW process, the heat input was principally provided by the friction between the two shoulders and surfaces of the workpiece, while the pin guaranteed the plastic flowing of the material inside. The positions of upper and lower shoulders were approximately symmetrical to the weld mid-thickness plane, resulting in a symmetrical heat generation and material flow around the middle layer of welds [6,27], and therefore, a relatively homogeneous distribution of the microstructure as well as mechanical properties along the thickness direction was produced.

3.4.2 Tensile properties

Fig. 10 presents the tensile test results of the whole and different slices of joints obtained under different welding conditions as well as the BM. It can be seen that the BM exhibited the highest yield strength (YS), ultimate tensile strength (UTS) and elongation with values of 443 MPa, 500 MPa and 18.7%, respectively. Specimens welded at the rotational speed of 300 rpm for SS-FSW showed approximately identical YS as those produced at 600 rpm except for the top slice of joint. And the YS of the top, middle and bottom slices of joints presented a decreasing trend with values of 318 MPa, 303 MPa and 274 MPa, respectively, the average of which was substantially the same as that of the whole joint. It was evident that the corresponding UTS, however, exhibited a slightly lower gradient along the thickness direction with values of 412 MPa, 399 MPa and 390 MPa, respectively, reaching the highest joint efficiency (UTS_{JOINT}/UTS_{BM}) of 82%, and the whole joint presented an approximate UTS of 410 MPa close to the one of top slice (412 MPa), as shown in Fig. 10c. When the rotational speed was increased to 600 rpm, the YS were 262 MPa, 316 MPa and 283 MPa across the thickness, while the UTS dropped to 273 MPa, 396 MPa and 386

MPa, respectively. The top slices of joint all fractured at the WNZ, resulting a worse YS and UTS. The elongation was measured by the average strain over the gauge length involving all the different regions. The highest elongation was obtained from the whole joint of SS-FSW at the lower rotational speed, with a value of ~10.2%. Compared with the slices along the thickness direction, the elongation of the whole joint increased significantly for SS-FSW at lower rotational speed, exhibiting a reverse tendency with the YS. These results may be related to the larger effective bearing area of the whole joint, which can support a higher ultimate tensile load than different slices of joint during the tensile process, since all the samples had similar UTS. And the same strain rate made the whole joint have experienced more sufficient plastic deformation. Additionally, the time of crack propagation and necking till the final fracture was relatively longer for the whole joint after the local material reached the critical value of fracture strength and microcracks formed, resulting in the whole joint owning an obviously higher elongation than slices of the joint. Since the samples welded at rotational speed of 300 rpm exhibited a higher microhardness distribution than those produced at 600 rpm (see Fig. 9), which can slow down the formation and growth rate of cracks, leading to a more enhanced ductility and UTS.

By comparison, tensile strengths of BB-FSW specimens degraded significantly as can be seen in Fig. 10. All samples had failed before the tensile stress reached YS. The UTS of the whole joint was 161 MPa at the rotational speed of 150 rpm, only achieved 32% of the BM, while different slices of joints presented relatively higher UTS with joint efficiencies of 44%, 42% and 47% for the top, middle and bottom slices, respectively. Furthermore, the elongation of BB-FSW joints decreased remarkably as shown in Fig. 10d. It has to be pointed that an increased rotational speed led to worse tensile strengths. The higher thermal cycle was mainly responsible for the difference in mechanical properties of two FSW techniques as mentioned in 3.1 and 3.2, which will also be discussed further below.

3.4.3 Fracture behaviors

Fig. 11 shows the evolution process of a heterogeneous strain distribution of the whole and different slices of SS-FSW samples, which were measured by DIC with the purpose of evaluating quantitatively the deformation of different zones during the tensile test [30,31]. And the YS distribution map of the whole specimen is presented in Fig. 12, where the YS values were calculated according to the strain evolution of adjacent grid points along the loading direction for every image, assuming that the local stress in every region equaled to the global stress.

Firstly, the strain was relatively homogeneous without the strain concentration through the transverse section before the stress reached 216 MPa, this stage was mainly dominated by elastic deformation. Then, the deformation localized in the HAZ of the RS at the stress of 246 MPa, where existed the minimum YS (see Fig. 12). It is known during the tensile process, the weakest region of the joint is more susceptible to the stress-strain concentration. When the YS reached 285 MPa, the strain response of the specimen was located in the HAZ of the RS principally and the AS partly, with less degree of deformation in TMAZ and WNZ. The local strain in the HAZ of the AS became more distinct gradually than that in the RS with the increase of stress, which was attributed to the plastic deformation strengthening in the RS, followed by a more rapid localization of strain in the HAZ of the AS consequently. At the strain state close to the fracture, the occurrence of necking was observed at the corresponding region. It can be expected the strain localization took place before the plastic deformation of joints, and finally the failure took place as about 45˚ shear fracture in the HAZ of the AS, where existed the lowest microhardness value (shown in Fig. 8) and it was more favorable for the crack initiation. Moreover, the local strain reached the maximum value of ~37% before fracture. The strain states at the frames of YS, UTS and failure for different slices are presented in Fig. 11b-d. It can be seen, the fracture of different slices was in the same process as the whole joint, while the degrees of deformation before failures were less than that of the whole sample and exhibited an increased trend along the thickness direction, with values of 15%, 17% and 22%, respectively, corresponding to the elongation trend (Fig. 10d). Moreover, as seen in Fig. 12, there was a moderate increase in YS of the WNZ along the thickness direction, which was in accordance with the evolution results of grain sizes (Fig. 7 and Fig. 8).

Fig. 13 shows the DIC results of strain distribution of BB-FSW samples produced at the lower rotational speed, including the states of UTS and failure of the whole and different slices of the joint. In contrast, the variation of strain for BB-FSW joints was entirely different, and the largest local strain was only with a value of ~1.1% when a rapid fracture happened suddenly, suggesting that the fracture location of the sample was still under elastic loading. It should be mentioned although the local softening regions led to the strain concentration in HAZ at the initial stage of deformation, for all BB-FSW specimens produced at different rotational speeds, the failure occurred in the WNZ without plastic deformation. It was evident that the location of crack initiation was not related to any softening regions in the WNZ, according to the Fig. 8d and 8e. When considering the microstructure characteristics of welds, however, the crack initiation location was exactly identical to the S line defect (shown in Fig. 5d), which was resulted from the higher peak temperature during welding operation, suggesting the S line had a negative effect on mechanical properties of joints, as the crack growth was mainly along the boundaries of S line, which was of much difference from the fracture behavior of SS-FSW samples. Also, it is reasonable to hypothesize that the same defect existed in the WNZ of BB-FSW joint welded at the lower rotational speed, without presenting obviously in the weld morphology actually (shown in Fig. 5c), and it was probably attributed to the insufficient etching treatment during the preparation of metallographic samples. Additionally, a more significant defect of S line was produced when the rotational speed increased, resulting in a more severe deterioration of mechanical properties in WNZ of the joint.

The representative SEM micrographs of the fracture surface of SS-FSW and BB-FSW samples are presented in Fig. 14. It was observed that the SS-FSW joint which showed higher elongation contained a fracture surface with a few small dimples as well as tearing ridges, indicating the similar characteristics as a ductile mechanism of fracture. For BB-FSW joints, there were no remnant particles on the fracture surface in spite of the fracture behavior related to the S line, signifying a brittle mode of fracture.

4. Conclusions

A 12 mm thick 7085-T7452 aluminum alloy has been conventional single side and bobbin tool friction stir welded, respectively, and the temperature distribution, microstructures and mechanical properties of the joints along the thickness direction was investigated in detail. The conclusions of significance are drawn as follows:

(1) The distribution of temperature gradient along the thickness direction was nonuniform for SS-FSW joints, exhibiting a decreased trend, while the temperature of lower shoulder affected zones was slightly higher than that of upper shoulder affected ones for the BB-FSW joints, introducing a much smaller temperature gradient along the thickness direction.

(2) The bowl-shaped cross section of SS-FSW joints and dumbbell-shaped cross section of BB-FSW joints were all divided into three different areas: WNZ, TMAZ and HAZ. The boundary between WNZ and TMAZ and the lines of plastic flow were more obvious in the AS of weld. There was no observed difference of grain size along the thickness direction at WNZ of BB-FSW welds.

(3) The microhardness distribution of both FSW joints was similar to W shape, and the lowest hardness locates at HAZ in AS. The UTS of the SS-FSW joints reached its peak value (410 MPa), at a rotational speed of 300 rpm and welding speed of 60 mm/min, which is 82% of the parental material, more than BB-FSW, whose joint efficiency only reached 47%. The whole SS-FSW joints present a much higher elongation than the slices. Moreover, the deformation localized in the HAZ of the RS at the stress of 246 MPa (minimum YS) and about 45˚ shear fracture is located in the

HAZ of the AS (lowest microhardness value) due to the rapid localization of strain.

(4) Failure initiates from HAZ for SS-FSW samples through ductile fracture, while the fracture mode changes to brittle fracture in WNZ for BB-FSW specimens.

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