Study on Nonhomogeneity of Low-Cycle Fatigue Properties Along the Thickness Direction of Thick-Plate Aluminum Alloy Friction Stir Welded Joints

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Abstract

Low-cycle fatigue experiments were conducted on layered slices of 2219-T62 aluminum alloy base material and friction stir welding (FSW) joints along the thickness direction, to investigate the influence of non-uniform thermo-mechanical coupling and welding process parameters on the low-cycle fatigue deformation behavior of the joints. The results indicate that: at low applied total strain amplitudes of 0.1% and 0.2%, the low-cycle fatigue cyclic stress amplitude and plastic strain amplitude curves exhibit a horizontal distribution, fatigue cracks initiate on the specimen surface with singularity, and the crack propagation zone shows distinct fatigue striations; at high strain amplitudes of 0.4%, 0.6%, and 0.8%, the stress amplitude increases, the plastic strain amplitude decreases, cyclic hardening occurs, and with increasing strain amplitude, the degree of fatigue cyclic strain hardening gradually increases, crack initiation exhibits multiple sources, and the fracture is dominated by dimple fracture characteristics. The fatigue life of the base material is higher than that of the FSW joint; compared with the middle and bottom slices of the joint, the upper slice exhibits lower fatigue life. With increasing rotational speed, the fatigue life decreases, while the welding speed has a minor influence on it.

Full Text

Study on Nonhomogeneity of Low-Cycle Fatigue Properties Along the Thickness Direction of Thick-Plate Aluminum Alloy Friction Stir Welded Joints

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Abstract Low-cycle fatigue tests were carried out on layered slices taken along the plate-thickness direction from 2219-T62 aluminum alloy base material and friction stir welded (FSW) joints, and the effects of the nonuniformity of thermomechanical coupling and welding-process parameters on the low-cycle fatigue deformation behavior of the joints were investigated. The results show that, at low applied total strain amplitudes of 0.1% and 0.2%, the cyclic stress-amplitude and plastic-strain-amplitude curves are horizontally distributed; fatigue cracks initiate at the specimen surface and are single in nature, and obvious fatigue striations appear in the crack-propagation region. At high strain amplitudes of 0.4%, 0.6%, and 0.8%, the stress amplitude increases, the plastic strain amplitude decreases, cyclic hardening occurs, and, as the strain amplitude increases, the degree of cyclic strain hardening gradually rises; crack initiation is characterized by multiple sources and is dominated by ductile fracture features. The fatigue life of the base material is higher than that of the FSW joint; compared with slices from the middle and bottom of the joint, the fatigue life of the upper slice is lower. As the rotational speed increases, the fatigue life decreases, while the welding speed has only a slight effect on it.

Keywords friction stir welding; thick-plate aluminum alloy; low-cycle fatigue properties; cyclic deformation; fracture mechanism

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STUDY ON NONHOMOGENEITY OF LOW-CYCLE FATIGUE PROPERTIES ALONG THICKNESS DIRECTION OF PLATE FOR FRICTION STIR WELDED ALUMINUM ALLOY JOINT

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ABSTRACT The low-cycle fatigue tests have been conducted for the base material and friction stir welded

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(FSW) joints along the thickness direction of 20 mm 2219-T62 aluminum alloy thick plate. The influence of coupled thermal–mechanical and welding parameters on the fatigue deformation behavior of the joint has also been systematically investigated. The results show that stress amplitudes and plastic strain amplitudes present horizontal distribution during cyclic deformation at low strain amplitudes of 0.1% and 0.2%. Moreover, single fatigue crack initiates from the surface or near-surface defects and propagation zone is basically characterized by fatigue striations. While at higher strain amplitudes of 0.4%, 0.6% and 0.8%, the stress amplitude increases and the plastic strain amplitude decreases as cyclic deformation proceeded, indicating that cyclic hardening occurred, and the degree of hardening increases. The fracture surface presents a multi-source initiation and dimple-like features. Based metal (BM) exhibits higher stress amplitude and low hysteresis energy at mid-life which causes less fatigue damage at the same strain amplitude, resulted in a longer fatigue lifetime. Compared to middle and bottom slices, the top has higher stress amplitude and lower plastic strain amplitude corresponding to a lower fatigue life. Moreover, both the propagation zone area in the top and bottom slice are basically the same, fatigue striation in the bottom increases. The higher the rotational rate, the higher the stress amplitude, the lower the plastic strain amplitude and fatigue life, but no strong effect of the welding speed on the fatigue life can be seen.

KEY WORDS friction stir weld, thick aluminum plate, low-cycle fatigue property, cyclic deformation, mechanism of fracture

During friction stir welding (FSW) of thick aluminum-alloy plates, the weld material experiences a temperature gradient that is high at the top and low at the bottom along the plate-thickness direction, as well as nonuniform plastic deformation. This leads to inhomogeneity in the microstructure and the morphology/distribution characteristics of second-phase particles in the weld, thereby causing large differences in the tensile properties of the joint. The tensile fracture also differs markedly along the plate-thickness direction: compared with the upper slice, which has higher tensile strength, the bottom slice has lower strength but higher elongation[1-4]. When the lower-strength bottom material of a thick-plate aluminum-alloy FSW joint is subjected to external force, it reaches the fracture strength first; the higher-strength upper slice of the joint, however, fractures without complete plastic deformation, resulting in a reduction in the overall strength of the joint[1,4]. Aluminum-alloy structural components are used in the aerospace industry owing to their excellent comprehensive mechanical properties. Under actual service conditions, however, components are subjected to high stresses, which cause plastic deformation in local regions and produce low-cycle fatigue, seriously affecting service life[5]. Therefore, extensive attention has been paid to the stability of aluminum-alloy properties and to low-cycle fatigue performance. Compared with tensile properties, the cyclic load peak that induces fatigue failure is often far smaller than the "safe" load estimated from static fracture analysis. Fatigue is one of the main pathways for failure of materials and components and is an important indicator for evaluating weld-joint quality. In actual production applications, failures of machinery and structural components are mostly caused by some fatigue process. To date, studies on the fatigue properties of aluminum-alloy FSW joints have mainly focused on the overall fatigue life of joints[6-9] and fatigue-crack propagation properties[10-13]. Among reports on the overall low-cycle fatigue properties of FSW joints, Czechowski[14] and Ceschini et al.[15,16] studied the low-cycle fatigue properties of integral FSW joints of aluminum alloys 6061/Al_2O_3p and 7005/Al_2O_3p, respectively. Feng et al.[17,18] studied the low-cycle fatigue life, cyclic deformation behavior, and fracture mechanism of 7075 and 6061 aluminum-alloy integral FSW joints. However, there have been few reports on fatigue-property studies of thick-plate aluminum-alloy FSW joints by layered slicing along the plate-thickness direction[19]. Therefore, it is necessary to investigate the variation law of fatigue properties of FSW joints along the plate-thickness direction, so as to provide theoretical support for engineering applications of thick-plate aluminum-alloy FSW.

In view of the present unclear understanding of the variation law of fatigue properties and fracture mechanisms of FSW joints along the plate-thickness direction, and of the fact that further research is urgently needed, this work focuses on studying the fatigue properties of layered slices of FSW joints along the plate-thickness direction under different applied total strain-amplitude conditions, including fatigue life, cyclic stress–strain response behavior, low-cycle fatigue parameters, and the microscopic morphology of failed fracture surfaces. The work reveals the variation law of fatigue properties of thick-plate aluminum-alloy FSW joints along the plate-thickness direction and explores damage and fracture mechanisms, thereby providing a theoretical basis for studying the welding-process mechanism of thick-plate aluminum-alloy FSW and for optimizing process parameters.

1 Experimental Methods

The experimental material was a 20 mm-thick 2219-T62 aluminum-alloy rolled plate, with the rolling direction being the length direction. The specimen size was 300 mm × 100 mm × 14 mm. Butt welding along the specimen length

Fig. 1

Figure 1: Fig. 1

direction was carried out on an FSW-2SLM-1040 thick-plate friction stir welding machine independently developed by the China Friction Stir Welding Center. The stirring tool used had a shoulder diameter of 34 mm, and a threaded conical pin with left-hand threads on its side surface, with a tip diameter of 9.3 mm and a root diameter of 13.5 mm. The pin length was 19.7 mm, the welding tilt angle was 2.8°, the rotation direction was clockwise, and the plunge depth was sufficient. FSW was performed using tool rotational speeds of 300, 400, and 500 r/min and welding speeds of 50, 80, and 100 mm/min. The cooling method was spraying toward the stirring tool and the upper surface of the weld.

with a constant flow rate.

After welding, specimens were cut from the transverse section of the FSW weldment (perpendicular to the welding direction). After rough grinding, fine grinding, and mechanical polishing, the specimen surface was etched with Keller’ s reagent (1 mL HF + 5 mL HNO_3 + 3 mL HCl + 190 mL H_2O). An OLYMPUS-PMG3 optical microscope (OM) was used to observe, in three layers along the plate-thickness direction (top, middle, and bottom), the microstructural evolution of the base metal and the weld nugget zone (WNZ), thermo-mechanically affected zone (TMAZ), and heat affected zone (HAZ) of the FSW weld. A JSM-6380LV scanning electron microscope (SEM) was used to observe the grain size and morphological characteristics of the second-phase-particle distribution.

According to the ASTM E8E standard, the specimens were machined by wire electrical discharge machining into the fatigue specimens shown in Fig. 1. Meanwhile, they were divided into three equal parts along the plate-thickness direction. No. 120, 240, 320, and 600 water-abrasive papers were selected to grind the cut machined surfaces along the loading-axis direction, so as to avoid surface stress concentration during experimental loading. The ground specimens were subjected to strain-controlled tension–compression fatigue tests on an Instron 8801 fatigue testing machine. For the low-cycle fatigue tests, $R = -1$ (R is the cyclic strain ratio) and a constant strain rate of 1×10^{-2} s^{-1} were selected, and the loading waveform was a constant-amplitude sinusoidal wave. Before 1×10^5 cyc cycles, the low-cycle fatigue tests were controlled by strain; the applied total strain amplitudes were 0.1%, 0.2%, 0.3%, 0.4%, 0.6%, and 0.8%, respectively. After 1×10^5 cyc cycles, if the specimen had not failed, the test was changed to stress control, with a loading frequency of 50 Hz. SEM observation was conducted on the fracture surfaces of the fatigue specimens to analyze their fracture mechanisms.

Fig.1 Schematic of configuration and size of fatigue specimens (The 20 mm thick tensile specimen is divided into 6.6 mm ones by linear cutting. Unit: mm)

2 Experimental Results and Discussion

2.1 Effect of Welding Parameters on the Microstructure of FSW Welds

Figure 2 shows OM images of the microstructures and SEM images of the second-phase-particle distribution in the base metal of 20 mm thick 2219 aluminum alloy, in the middle-section HAZ of the FSW weld, and in the top-section TMAZ. As can be seen from Fig. 2a, the base-metal region consists of relatively large elongated plate-like structures, with widths and lengths of approximately 150 and 500 μm, respectively; the second-phase particles are distributed in a network along the initial grain boundaries. The HAZ is affected by the welding thermal cycle, but is not affected by mechanical stirring; the influence of the welding thermal cycle is weaker than that in the WNZ and TMAZ. In fact, the HAZ has undergone a special annealing-treatment process. Compared with the base metal, the metal undergoes very little distortion, but obvious grain growth occurs in the HAZ, and the precipitates within the grains are also markedly coarsened, as shown in Fig. 2b. The TMAZ likewise undergoes the combined action of high temperature and deformation, except that the deformation is not as severe as in the weld nugget zone. In this region, under the mechanical stirring action of the shoulder and the pin, fragmentation and adhesion/growth occur in the region near the weld nugget zone. In other regions, recovery and recrystallization occur under the welding thermal cycle; however, the original fibrous structure is subjected to the viscous shear force generated when the metal of the softened layer flows, resulting in a certain degree of elongation and bending deformation of the grains along the force direction, as shown in Fig. 2c.

Figure 3 presents OM images of the microstructures and SEM images of the second-phase-particle distribution at different positions through the plate thickness in the WNZ of water-cooled FSW welds at different rotational speeds of 300 and 500 r/min. Compared with the middle (Figs. 3b and e) and bottom (Figs. 3c and f) of the WNZ, the recrystallized grains in the top region (Figs. 3a and d) are larger. Compared with 300 r/min (Figs. 3a–c), at a rotational speed of 500 r/min (Figs. 3d–f), no obvious coarsening occurs in the grains at the top of the WNZ, whereas the recrystallized grains in the middle and bottom regions grow; the second-phase particles in the middle of the WNZ are obviously coarsened, while those in the top and bottom regions are refined. Because the upper surface of the weld is subjected to water cooling, growth of the recrystallized grains in the upper part of the weld is inhibited by the water-cooling effect, and the rotational speed has little influence on the grain size in the upper part of the water-cooled WNZ. At a high rotational speed (500 r/min), the metal in the middle of the weld is subjected to the thermal-cycle action of a higher peak temperature and a longer time; during the FSW process, plas–

Fig.2 OM images of microstructures and SEM images of second-phase particles (insets) in based metal (BM) of 2219 aluminum alloy (a), middle heat affected

Fig. 2

Figure 2: Fig. 2

Fig. 3

Figure 3: Fig. 3

zone (HAZ) (b) and top thermo-mechanical affected zone (TMAZ) (c) in friction stir welded (FSWed) joint

Fig. 3 OM images of microstructures and SEM images of second-phase particles (insets) in weld nugget zone (WNZ) at top (a, d), middle (b, e) and bottom (c, f) with water cooling under rotational rates of 300 r/min (a~c) and 500 r/min (d~f)

the material may undergo abnormal plastic flow, and the second-phase particles grow abnormally (Fig. 3e). When the rotational speed is relatively high, the material in the upper and bottom regions of the weld is subjected to large and sufficient mechanical stirring; the second-phase particles are broken up, forming a uniformly fine dispersion of second phases (Fig. 3d and f).

2.2 Fatigue life

Figure 4 shows the fatigue lives of the 20 mm-thick 2219 aluminum alloy base metal and the layered slices along the plate-thickness direction of water-cooled FSW joints under different applied total strain amplitudes and welding parameters. It can be seen that the larger the applied total strain amplitude, the lower the fatigue life; the fatigue life in the upper part of the weld is lower than that in the middle and bottom parts, especially at high strain amplitudes of 0.8%, 0.6%, and 0.4%. As can be seen from Fig. 4a, the fatigue lives of base-metal slices from different layers along the plate-thickness direction remain basically unchanged, indicating that wire cutting has almost no influence on the layered cutting of the thick plate, i.e., it is feasible to study the fatigue properties of the material along the plate-thickness direction by the wire-cutting method. From Fig. 4b~d, it can be seen that, when the rotational speed increases, the fatigue life of the joint decreases. This is because, at the high rotational speed of 500 r/min, grain growth and coarsening of the second phase occur (Fig. 3); under the cyclic tensile–compressive stress, cracking readily initiates at the contact between the relatively large second phases and the matrix, and the crack-propagation resistance is small, so cracks propagate easily and the fatigue life decreases. From Fig. 4e and f, it can be seen that the effect of welding speed on the fatigue life of layered slices of the FSW joint is relatively small, consistent with the tensile properties; this is caused by the relatively small influence of welding speed on the microstructure of the joint.

2.3 Cyclic stress-strain response behavior

Figure 5 shows the cyclic stress-strain response curves of layered slices of 20 mm-thick 2219 aluminum alloy water-cooled FSW joints at a welding speed of 80 mm/min and rotational speeds of 300 and 500 r/min (the applied total strain amplitude decreases gradually from 0.8% at the top of the figure to 0.1% at the bottom). It can be seen from Fig. 5 that the 2219 aluminum alloy base metal has relatively high cyclic stress amplitude and tensile strength (load-bearing capacity; strain rate 1×10^{-2} s^{-1}); during short-time fatigue deformation, the accumulated cyclic stress is difficult to reach the fracture limit of the material, and therefore the fatigue life is relatively long. When the applied total strain amplitude is relatively low (0.1%), stable cyclic stress and strain response behavior is exhibited throughout the cyclic deformation period. When the applied total strain amplitudes are 0.2% and 0.3%, cyclic stability is observed in the initial stage of fatigue deformation, followed by an increase in cyclic stress and a decrease in cyclic strain amplitude, and the material exhibits cyclic strain hardening. When the applied total strain amplitudes are 0.8%, 0.6%, and 0.4%, the cyclic stress shows an increasing trend throughout the cyclic deformation period, while the cyclic plastic strain amplitude decreases until specimen fracture; in particular, at 0.8% and 0.6%, the specimens undergoing fatigue cyclic deformation exhibit obvious cyclic strain-hardening behavior. The cyclic hardening characteristics in the upper part of the FSW joint are significantly stronger than those in the bottom part as the applied total strain amplitude increases. This is because, compared with the bottom part, the fine, dispersed second-phase particles in the upper part of the weld exert a stronger impediment to dislocation motion, and the coarse grains have greater dislocation storage capacity; therefore, the cyclic hardening in the upper part of the joint is higher. Compared with the middle and bottom slices of the FSW joint, the upper slices exhibit higher cyclic stress and lower cyclic plastic strain during cyclic deformation, resulting in a lower low-cycle fatigue life. Meanwhile, it can be seen from Fig. 5 that, under the condition of high rotational speed (500 r/min), the cyclic stress of the layered slices of the FSW joint is significantly higher than that at the low rotational speed (300 r/min), while the cyclic plastic strain amplitude and fatigue life are lower. This is because, in high-rotational-speed FSW welds, the grains and second-phase particles grow (Fig. 3), resulting in a decrease in fatigue life.

The cyclic stress response behavior is controlled by microstructural stability, second-phase particles, dislocation multiplication, and slip systems. The cyclic hardening in Fig. 5 is mainly because a large number of dislocations are generated inside the material during cyclic deformation. These high-density dislocations interact with one another during motion, forming complex dislocation configurations such as dislocation tangles and Lomer–Cottrell sessile dislocations, all of which hinder further dislocation ⋯

Fig.4 Fatigue life vs total strain amplitude ($\Delta\varepsilon_t$) of BM and the slices of FSW joints with water cooling along thickness direction of plate under different rota-

Figure 4. Fatigue life vs total strain amplitude ($\Delta\varepsilon_t$) of BM and the slices of FSW joints with water cooling along the thickness direction of the plate under different rotational rates and welding speeds (N_f—number of cycles to failure). (a) BM; (b) 300 r/min, 80 mm/min; (c) 400 r/min, 80 mm/min; (d) 500 r/min, 80 mm/min; (e) 300 r/min, 50 mm/min; (f) 300 r/min, 100 mm/min.

Figure 4: Figure 4. Fatigue life vs total strain amplitude ($\Delta\varepsilon_t$) of BM and the slices of FSW joints with water cooling along the thickness direction of the plate under different rotational rates and welding speeds (N_f—number of cycles to failure). (a) BM; (b) 300 r/min, 80 mm/min; (c) 400 r/min, 80 mm/min; (d) 500 r/min, 80 mm/min; (e) 300 r/min, 50 mm/min; (f) 300 r/min, 100 mm/min.

tional rates and welding speeds (N_f—number of cycle to failure)
(a) BM (b) 300 r/min, 80 mm/min (c) 400 r/min, 80 mm/min
(d) 500 r/min, 80 mm/min (e) 300 r/min, 50 mm/min (f) 300 r/min, 100 mm/min

dislocation motion and reduces dislocation mobility, thereby causing cyclic hardening[^17,^18]. In addition, dislocation motion mainly occurs in the lower-strength α-Al matrix phase, while the high-strength second-phase particles Al_2Cu distributed on the matrix are the main obstacles to dislocation motion (Figs. 2 and 3). Dislocations can only bypass these particles through the Orowan mechanism or pile up in front of them; as a result, local hardening occurs on the slip plane, further impeding dislocation motion and producing dispersion strengthening[^17,^18,^20]. At the same time, the larger the applied total strain amplitude, the greater the stress required for fatigue deformation of the material; the number of grains capable of undergoing plastic deformation increases, and the number of slip systems that can be activated in each grain also increases. Therefore, the probability of interactions among dislocations, and between dislocations and precipitates, increases, enhancing the degree of cyclic strain hardening. Thus, the greater the applied total strain amplitude, the higher the degree of cyclic strain hardening. While the material is continuously strengthened by dislocation multiplication, dislocation annihilation also occurs: dislocations of opposite sign meet during dislocation motion and cancel each other out. As a result, the resistance to dislocation motion during fatigue plastic deformation decreases, and the applied stress required for dislocation motion is reduced, producing the so-called softening effect. For fatigue deformation at relatively low applied total strain amplitudes (0.1% and 0.2%), dislocation multiplication is relatively slow, and dislocation multiplication and annihilation readily reach equilibrium, so that the hardening and softening effects of the material cancel each other out. Consequently, the cyclic stress amplitude and plastic strain amplitude remain essentially unchanged during cyclic deformation.

Figure 6 shows the cyclic stress and strain of the FSW layered slices under the conditions of a rotational speed of 300 r/min and welding speeds of 50 and 100 mm/min.

Fig. 5: Cyclic stress amplitude and plastic strain amplitude vs fatigue life under rotational rates of 300 and 500 r/min

Figure 5: Fig. 5: Cyclic stress amplitude and plastic strain amplitude vs fatigue life under rotational rates of 300 and 500 r/min

Fig. 6: Cyclic stress amplitude and plastic strain amplitude vs N under welding speeds of 50 and 100 mm/min

Figure 6: Fig. 6: Cyclic stress amplitude and plastic strain amplitude vs N under welding speeds of 50 and 100 mm/min

Fig. 5 Cyclic stress amplitude (a, c) and plastic strain amplitude (b, d) vs fatigue life (N) of BM and slices of FSW joints along the thickness of the plate under rotational rates of 300 r/min (a, b) and 500 r/min (c, d), respectively.

Panel labels and legends visible in Fig. 5:
(a) Stress amplitude / MPa vs N / cyc; legend: BM, Top, Middle, Bottom; strain range marked from 0.8% to 0.1%.
(b) Plastic strain amplitude / % vs N / cyc; legend: BM, Top, Middle, Bottom; strain range marked from 0.8% to 0.1%.
(c) Stress amplitude / MPa vs N / cyc; legend: Top, Middle, Bottom; strain range marked from 0.8% to 0.1%.
(d) Plastic strain amplitude / % vs N / cyc; legend: Top, Middle, Bottom; strain range marked from 0.8% to 0.1%.

Fig. 6 Cyclic stress amplitude (a, c) and plastic strain amplitude (b, d) vs N of BM and slices of FSW joints along the thickness of the plate under welding speeds of 50 mm/min (a, b) and 100 mm/min (c, d), respectively.

Panel labels and legends visible in Fig. 6:
(a) Stress amplitude / MPa vs N / cyc; legend: Top, Middle, Bottom; strain range marked from 0.8% to 0.1%.
(b) Plastic strain amplitude / % vs N / cyc; legend: Top, Middle, Bottom; strain range marked from 0.8% to 0.1%.
(c) Stress amplitude / MPa vs N / cyc; legend: Top, Middle, Bottom; strain range marked from 0.8% to 0.1%.
(d) Plastic strain amplitude / % vs N / cyc; legend: Top, Middle, Bottom; strain range marked from 0.8% to 0.1%.

response curves. Compared with a welding speed of 100 mm/min (Fig. 6c and d), at a welding speed of 50 mm/min (Fig. 6a and b), the fatigue cyclic stress amplitude of the layered slices of the FSW joint increases, the plastic strain amplitude decreases, and the fatigue life decreases.

2.4 Low-cycle fatigue parameters

In low-cycle fatigue tests, the cyclic stress–strain curve of a material reflects the true stress-strain characteristics of the material under low-cycle fatigue loading. The relationship between cyclic stress and strain of the material can be expressed by the following power law equation

18

, namely:

$$\frac{\Delta\sigma}{2} = K' \left(\frac{\Delta\varepsilon_{\mathrm{p}}}{2}\right)^{n'} \tag{1}$$

where $\Delta\sigma/2$ is the cyclic stress amplitude, $\Delta\varepsilon_{\mathrm{p}}/2$ is the plastic strain amplitude, K' is the cyclic strength coefficient, and n' is the cyclic strain-hardening exponent. Both $\Delta\sigma/2$ and $\Delta\varepsilon_{\mathrm{p}}/2$ are obtained from the cyclic hysteresis loops at half-life under different strain amplitudes. Linear regression analysis was performed on the experimental data, and K' and n' were calculated according to Eq. (1); the specific calculated values of these two parameters are shown in Table 1. It can be seen that, in terms of fatigue K' and n', the 2219 aluminum alloy base metal is significantly higher than the FSW joint, and the upper slice of the water-cooled FSW joint is higher than the bottom slice, indicating that the fatigue properties of the FSW joint are inhomogeneous along the plate thickness direction.

According to the stress-amplitude and strain-amplitude data obtained from the cyclic hysteresis loops at half-life, the cyclic stress–strain curves of the 2219 aluminum alloy base metal and of the slices at different layers through the thickness of the FSW joint were obtained. At 0.2% cyclic strain, the cyclic stress value corresponding to the intersection of the line parallel to the elastic deformation line and the cyclic stress–strain curve is referred to as the cyclic yield strength σ_y'. For strain-controlled low-cycle fatigue of the base metal and of the different layered slices of the FSW joint, σ_y' exhibits the same trend as fatigue K' and n'.

It is known from the literature

18, 21

that, for total-strain-controlled low-cycle fatigue tests, the total strain amplitude $\Delta\varepsilon_{\mathrm{t}}/2$ consists of the elastic strain amplitude $\Delta\varepsilon_{\mathrm{e}}/2$ and the plastic strain amplitude $\Delta\varepsilon_{\mathrm{p}}/2$, namely:

$$\left(\frac{\Delta\varepsilon_{\mathrm{t}}}{2}\right) = \left(\frac{\Delta\varepsilon_{\mathrm{e}}}{2}\right) + \left(\frac{\Delta\varepsilon_{\mathrm{p}}}{2}\right) \tag{2}$$

The elastic strain amplitude and plastic strain amplitude satisfy the following relationships with the fatigue life N_f, respectively

18

:

$$\left(\frac{\Delta\varepsilon_\mathrm{e}}{2}\right) = \frac{\sigma_\mathrm{f}'(2N_\mathrm{f})^b}{E} \tag{3}$$

$$\left(\frac{\Delta\varepsilon_\mathrm{p}}{2}\right) = \varepsilon_\mathrm{f}'(2N_\mathrm{f})^c \tag{4}$$

where σ_f' is the fatigue strength coefficient, b is the fatigue strength exponent, $2N_\mathrm{f}$ is the number of load reversals at fatigue failure (i.e., twice the fatigue life), E is the elastic modulus, ε_f' is the fatigue ductility coefficient, and c is the fatigue ductility exponent. Equation (3) is obtained by introducing the elastic modulus into the Basquin relationship, while Eq. (4) is the well-known Coffin-Manson relationship. Therefore, the relationship between total strain amplitude and fatigue life is expressed by Eq. (2) as

18

:

$$\left(\frac{\Delta\varepsilon_\mathrm{t}}{2}\right) = \left(\frac{\sigma_\mathrm{f}'(2N_\mathrm{f})^b}{E}\right) + \varepsilon_\mathrm{f}'(2N_\mathrm{f})^c \tag{5}$$

The obtained experimental results show that the fatigue deformation follows the modified Basquin and Coffin-Manson relationships. Based on Eqs. (3) and (4), using linear regression analysis, the low-cycle fatigue parameters of the 2219 aluminum alloy base metal and the slices at different layers of the water-cooled FSW joint can be obtained, as shown in Table 1. It can be seen that the base metal has the highest σ_f', and the lowest b and c. For the slices of the water-cooled FSW joint, from the top to the bottom along the plate thickness direction, σ_f' and c decrease, while b and ε_f' increase, which is basically consistent with the variation trend of fatigue life.

2.5 Fatigue fracture morphology

Figure 7a shows the layered microhardness distribution on the transverse cross-section of the layered slices of the FSW joint. It can be seen that the layered microhardness on the weld transverse cross-section exhibits a high-low-relatively high-low-high "W"-shaped distribution trend. The hardness values in the base-metal regions on both sides are the highest, decrease in the HAZ and TMAZ, and the lowest hardness zone (LHZ) in the upper part is located in the TMAZ, whereas the LHZs in the middle and bottom parts are located in the HAZ. The microhardness in the WNZ increases; the microhardness in the upper part of the WNZ is basically the same as that of the base metal, whereas that in the middle and bottom parts is lower. This is related to the grain size and second-

Table 1 Strain fatigue parameters of BM and FSWed joint slices along thickness direction of plate under rotational rate of 300 r/min and welding speed of 100 mm/min

Specimen	K'	n'	σ_y' / MPa	σ_f' / MPa	b	ε_f'	c	E / GPa
BM	893	0.1480	325	660	−0.1152	0.0640	−0.6909	71.9
FSW-Top	581	0.1059	283	539	−0.0929	0.0484	−0.6197	71.6
FSW-Bottom	559	0.0991	258	519	−0.0856	0.0764	−0.6207	76.3

Note: K'—cyclic strength coefficient, n'—cyclic strain hardening exponent, σ_y'—cyclic yield strength, σ_f'—fatigue strength coefficient, b—fatigue strength exponent, ε_f'—fatigue ductility coefficient, c—fatigue ductility exponent, E—Young' s modulus.

···consistent with the morphological characteristics of the particle distribution. As can be seen from the macroscopic fracture morphology of the weld in Fig. 7b, the fatigue-crack initiation source is located in the weld LHZ (the upper part of the weld is located in the TMAZ, while the middle and bottom are located in the HAZ). During cyclic deformation, the plastic deformation capacity and mutual coordination of the coarse grains in the softened zone are markedly reduced, and strong stress concentration is readily formed around the second-phase particles, leading to fracture; moreover, some second-phase particles near the fracture surface are fractured. The coarse grains and second-phase particles in the TMAZ at the upper part of the weld and in the HAZ at the middle and bottom (Fig. 2b and c) reduce the microhardness and critical fracture strength of this region. Therefore, during fatigue deformation of the layered slices of the FSW joint, macroscopic fracture occurs in the TMAZ at the upper part and in the HAZ at the middle and bottom.

Figure 8 shows SEM images of the low-cycle fatigue fracture surfaces of the 2219 aluminum-alloy base material under imposed total strain amplitudes of 0.2%, 0.4%, and 0.8%. At a low strain amplitude (0.2%), the fatigue fracture surface shows a fatigue-crack initiation source, a crack-propagation region, and a final rapid-fracture region. The fatigue crack initiates at the specimen surface, and obvious fatigue striations can be observed in the crack-propagation region (Fig. 8a). At high strain amplitudes, it is difficult to observe the fatigue-crack initiation source and crack-propagation region; instead, a typical ductile fracture morphology is exhibited (Fig. 8b and c). This may be caused by decohesion

Fig. 7

Figure 7: Fig. 7

along the interfaces between second-phase particles and grains, by fracture along the boundary line between a separated reticular second phase and the grains, or by direct intergranular fracture, and by secondary cracking of the second-phase particles. Surface defects on fatigue specimens act as notches, readily inducing stress concentration and promoting fatigue-crack initiation. This is the result of local shear stress: under the maximum shear stress, dislocation migration in the material surface layer forms fine slip bands; the greater the fatigue stress, the larger the number of slip bands. Under repeated fatigue tension–compression loading, reverse slip occurs on adjacent slip planes, and fatigue slip bands form grooves and intrusions on the material surface, ultimately leading to fatigue-crack initiation[^19,^20]. During fatigue deformation, once a fatigue crack initiates, it first propagates inward along the slip plane that is most consistent with the direction of the maximum shear stress (at 45° to the principal stress direction). This stage is the first stage of crack propagation. Because the crack extension per cycle in the first stage is very small, there are no pronounced microscopic features on the fracture surface, which resembles a cleavage morphology and shows no plastic-deformation features. Owing to the different orientations of the grains and the hindering effect of grain boundaries, after the crack has extended to a certain length, the plane of the principal stress at the crack tip deviates; fatigue-crack propagation gradually turns toward the direction perpendicular to the maximum normal stress, i.e., macroscopic crack propagation begins. This continues until the unfractured portion can no longer bear the applied load, at which point the crack begins unstable propagation; this is referred to as the second stage of fatigue-crack propagation. In the second stage of crack propagation, the crack propagates transgranularly and at a relatively high rate; in each stress cycle it advances by approximately a micrometer-scale distance. The fracture surface after the second stage of fatigue-crack propagation is mainly characterized by fatigue striations[^20].

Fig. 7 Distributions of microhardness along the transverse section of slice in FSW joint (a) and typical macrograph images showing the fracture features of samples (b)

Figure 9 shows SEM images of the low-cycle fatigue fracture surfaces of the upper slice of the water-cooled FSW joint under imposed total strain amplitudes of 0.2%, 0.4%, and 0.6%. It can be seen that, at a low strain amplitude (0.2%), fatigue-crack initiation has a single character (Fig. 9a), and the crack-propagation region is relatively large. Its microscopic feature is fatigue striations perpendicular to the crack-propagation direction. The final rapid-fracture region is relatively small, indicating ductile fracture; its main microscopic feature is a dimple morphology. Under cyclic tension–compression stress, during micro-zone plastic deformation in the material,

Fig. 8

Figure 8: Fig. 8

Fig. 9

Figure 9: Fig. 9

Fig. 8 SEM images of BM of 2219 aluminum alloy at strain amplitudes of 0.2% (a), 0.4% (b) and 0.6% (c) (Insets show the high magnified images corresponding to rectangles in Figs. 8a~c, respectively)

Fig. 9 SEM images of top slice of FSW joint at different strain amplitudes 0.2% (a), 0.4% (b) and 0.6% (c), and high magnified SEM images of zone A in Fig.9a (d), zone B in Fig.9b (e) and zone C in Fig.9c (f) (Inset in Fig.9d shows the higher magnified image)

Micropores are generated, then nucleate, grow, and coalesce, eventually connecting with one another and leading to fracture of the material, leaving traces on the fracture surface. At a high strain amplitude (0.4%), fatigue-crack initiation exhibits multiple origins (Fig. 9b), that is, fatigue cracks initiate at multiple locations on the specimen surface; the crack-propagation region is relatively small, and the rapid-fracture region occupies more than 80% of the entire fracture surface. At a higher strain amplitude (0.6%), no crack initiation or propagation occurs; only rapid fracture takes place, similar to tensile fracture, and the overall microscopic fracture surface shows typical dimple features (Fig. 9f). During low-cycle fatigue, when the stress concentration is not pronounced, a single fatigue core is often formed. For example, at a low strain amplitude, fatigue stress concentration is not readily produced, so a single fatigue-crack initiation source is formed (Fig. 9a). When a large stress concentration occurs, multiple-origin fracture is prone to occur. On the one hand, the specimen surface is in a plane-stress state, which facilitates plastic slip and the formation of extrusion and intrusion slip bands, thereby forming nuclei for microcracks. In addition, machining marks left on the specimen surface cause stress concentration at many weak surface locations under cyclic loading at a relatively large strain amplitude (0.4%), thereby initiating fatigue cracks and forming multiple-origin fatigue fracture as shown in Fig. 9b. During cyclic plastic deformation at a high strain amplitude (0.6%), stress concentration occurs and the critical fracture stress of the material is rapidly reached; crack propagation essentially has no time to release the stress before fracture has already occurred. During fatigue deformation, when a relatively small externally applied total strain amplitude is imposed, the fracture surface exhibits relatively well-developed fatigue-crack propagation features; the fatigue region is larger, the instantaneous-fracture region is smaller, and fatigue life is longer. Conversely, when a higher externally applied total strain amplitude is imposed, crack propagation is insufficient, the instantaneous-fracture area becomes larger, and fatigue life decreases.

3 Conclusions

(1) At low externally applied total strain amplitudes (0.1% and 0.2%), the low-cycle fatigue cyclic stress amplitude and plastic strain amplitude are horizontally distributed. At high strain amplitudes (0.4%, 0.6%, and 0.8%), the fatigue cyclic stress amplitude increases; the increase in amplitude is greatest at 0.8%, while the plastic strain amplitude decreases, indicating the occurrence of cyclic hardening. Moreover, as the strain amplitude increases, the degree of fatigue cyclic strain hardening gradually increases, fatigue damage increases, and fatigue life decreases. The stress amplitude of the base material is higher than that of the FSW joint. Compared with the middle and bottom slices of the joint, the upper slice has a higher cyclic stress amplitude and a lower cyclic plastic strain amplitude; its cyclic hardening characteristics become significantly stronger than those of the bottom slice as the strain amplitude increases, and its fatigue life is lower. As the rotation speed increases, the cyclic stress amplitude increases, the cyclic plastic strain amplitude decreases, and the fatigue life of the joint decreases. The welding speed has a relatively small effect on the fatigue life of the layered slices of the joint.

(2) The fatigue cyclic strength coefficient (K'), cyclic strain-hardening exponent (n'), and cyclic yield strength (σ'_y) of the 2219 aluminum-alloy base material are significantly higher than those of the FSW joint. Along the plate-thickness direction of the water-cooled FSW joint, these values gradually decrease from the upper slice to the bottom slice, indicating inhomogeneity of fatigue properties through the thickness direction of the thick-plate aluminum-alloy FSW joint.

(3) At a low externally applied strain amplitude (0.2%), fatigue cracks initiate on the specimen surface and are single-origin in nature. On the fatigue fracture surface, the fatigue-crack initiation source, crack-propagation region, and rapid-fracture region can be observed; the crack-propagation region contains obvious fatigue striations, the rapid-fracture region is relatively small, and the microscopic morphology is dominated by dimple features. At a high strain amplitude (0.4%), fatigue-crack initiation is multi-origin; the crack-propagation region is small, and the rapid-fracture region occupies more than 80% of the entire fracture surface. At a strain amplitude of 0.6%, no crack initiation or propagation occurs; only rapid fracture takes place, and the overall microscopic fracture surface shows typical dimple features.

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Figure 6

Figure 10: Figure 6

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Figures

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