Characterization and evaluation of stable localized corrosion in a 7075-T6 aluminum alloy FSW joint before and after anodizing treatment

Kai Li \(^1\), Hong Yang \(^2\), Mengting Zou \(^2\), Bingyuan Yang \(^2\), Huibin Xu \(^2\), Huaxia Zhao \(^1\), Henggang Yin \(^3\) and Yanlong Ma \(^*\)

\(^1\) AVIC Manufacturing Technology Institute, Beijing 100024, People’s Republic of China
\(^2\) College of Materials Science and Engineering, Chongqing University of Technology, Chongqing 400054, People’s Republic of China
\(^3\) Jiangsu Jingchuan Materials Testing and Research Co., Ltd, Suzhou 215124, People’s Republic of China

* Authors to whom any correspondence should be addressed.
E-mail: likai.85@foxmail.com and myl@cqut.edu.cn

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Abstract
The microstructure and corrosion processes of a friction stir welded (FSW) 7075-T6 aluminum alloy joint, before and after anodizing surface treatment, have been characterized by advanced techniques, and the feasibility of anodizing treatment as a corrosion mitigation method has been evaluated. The results showed that different zones of the FSW joint had distinctly different microstructure and consequently different corrosion behavior in NaCl solution. Stable localized corrosion occurred in the transitional regions between the thermo-mechanically affected zone (TMAZ) and heat affected zone (HAZ), and was characterized by intergranular corrosion. The intergranular corrosion was ascribed to the galvanic coupling effect between Cu-rich grain boundary precipitates and the precipitates free zones (PFZs). Although anodizing and the subsequent sealing treatments could greatly improve the corrosion resistance of the base metal, the TMAZ/HAZ transition regions still showed much higher corrosion susceptibility than other regions. The high corrosion susceptibility of the FSW joint after anodizing treatment is not ascribed to the difference of the anodic oxide film in the regions, but the heterogeneous microstructure of the alloy beneath the anodic film. The present paper has shown that the stable localized corrosion in the FSW joint is intrinsically stemmed from the welding process itself and traditional mitigation method such as anodizing treatment cannot solve the problem; more effective corrosion mitigation methods are still awaited.

1. Introduction
Aluminum alloys are extensively used in aircraft structures. Generally, 80% of a typical transport/cargo airframe is made of aluminum alloys. During the design of a new airframe, the emphasis is always put on mechanical life for which a period of approximately 20 years is planned. Nevertheless, beyond those 20 years of service, the aircraft is qualified of aging as their mechanical problems (such as fatigue) become somehow less important than the problems related to corrosion of aluminum structural parts [1, 2]. This is especially the case when aircraft operational environments are chemically aggressive, e.g., industrially polluted and seacoast atmospheres [3, 4]. In such environments, corrosion can occur and provoke stress corrosion cracking (SCC) when a suitable tensile stress applied to the parts. This is a major concern for the 7075 alloy, which presents excellent mechanical properties, but is susceptible to localized corrosion attack [5]. Moreover, the situation is further exacerbated by the presence of a weld seam [6].

One of the most widely used welding techniques for 7xxx Al alloys is friction stir welding (FSW) [7]. FSW joints usually exhibit several characteristic microstructure zones, namely weld nugget zone (NZ), thermo-mechanically affected zone (TMAZ) and heat affected zone (HAZ). Although FSW leads to 30 ~ 50% increase in...
joint strength compared with conventional welding techniques, it causes corrosion problems due to the complicated microstructural changes in the joint because of the plastic deformation and thermal cycles during FSW procedure [8, 9]. It is now generally believed that corrosion of joint is one of the major concerns for the application of FSW technology to aircraft structures [10, 11].

Much work has been carried out worldwide on corrosion mechanism of FSW aluminum alloy joints, and it is found that the corrosion performance and associated mechanism varies greatly for different alloys [12–25]. Frankel et al [26] investigated the susceptibility of welded and unwelded samples of Al 5454 in the −O and −H34 tempers to pitting corrosion and SCC in chloride solutions. It was found that the FSW samples exhibited superior resistance to pitting corrosion and adequate resistance to SCC compared to the base metal. Similarly, Shen et al [27] reported that the 5083 FSW joint showed a higher corrosion resistance in EXCO solution and a lower pitting tendency than the base alloy. Kang et al [28] investigated the corrosion behavior of an AA2024-T3 aluminum alloy FSW joint by using an in situ observation method, revealing that the density and degree of the pitting corrosion in the shoulder active zone were slightly larger compared to other regions on the top surface. Bousquet et al [29] systematically studied the relationship between microstructure, microhardness, and corrosion sensitivity of an AA 2024-T3 FSW joint, finding that the HAZ close to the TMAZ was the zone most sensitive to corrosion. Wadeson et al [30] studied the corrosion behavior of AA7108-T79 aluminum alloy FSW joint. The results revealed that the edge regions of the TMAZ were most susceptible to corrosion. The localized corrosion occurred inter-granularly due to the non-uniform distribution of η/η′ (MgZn2) precipitates within the TMAZ.

Recently, Zhang et al [31] reported similar corrosion phenomenon as Wadeson et al in a 2A97 Al-Cu-Li alloy FSW joint. They found that the welding process promoted the formation of a high population of T1 phase (Al6CuLi) precipitates in TMAZ, which consequently became susceptible to localized corrosion. Localized corrosion in TMAZ initially occurred at high angle grain boundaries, which were decorated by T1 phase precipitates and, subsequently, developed into grain interior through preferential dissolution of T1 phase precipitates.

It was suggested by the literature that, for high strength aluminum alloys such 2xxx and 7xxx, the FSW joint is usually more susceptible to localized corrosion than the base metal. Thus, the mitigation of the localized corrosion in FSW joints of these alloys is critical. A variant of FSW process, known as stationary shoulder FSW (SSFSW) was developed at TWI in 2005, which offers advantages in terms of improved weld profile, better control of weld microstructure, more relaxed requirements for joint fit-up and further reduction of weld flaws and distortion. The heat input and thermal-mechanical behavior of the new processes are different from the conventional FSW process, leading to the changes in the microstructure and properties of the joint [32]. Recently, it is also suggested that friction stir vibration welding (FSVW) could result in faster nucleation and growth of grains and thus finer grain sizes compared with conventional FSW, which improved the corrosion resistance of the weld joint [33]. However, in most cases, it is difficult to make a balance between mechanical properties and corrosion resistance. In industry, post-weld heat treatment, shot peening, laser cladding, micro arc-oxidation and anodizing are often used to improve the corrosion resistance of FSW joints. Of them, anodizing is the most common corrosion mitigation method for aluminum alloys [34, 35]. In recent years, as an alternative to the traditional chromic acid anodizing (CAA), tartaric-sulfuric acid anodizing (TSA) has been successfully applied in the aerospace industry. It is found that the addition of tartaric acid to sulfuric acid can alleviate the dissolution of oxide film in electrolyte and improve the density and therefore the corrosion resistance of the resultant anodic film [36, 37]. Further, it is reported that the corrosion resistance of the TSA film can be greatly improved by sealing with layered double hydroxides (LDH) [38–40].

In the present paper, the microstructure of a FSW 7075-T6 aluminum alloy joint was carefully characterized in the first place. Then, the corrosion behavior of the joint in chloride-containing environment was investigated, with particular focus on the effect of the alloy microstructure. Finally, the FSW joint was anodized in a TSA bath and then sealed with LDH to improve its corrosion resistance. The present work suggested that FSW joint of the 7075-T6 alloy is highly susceptible to stable localized corrosion, and traditional mitigation method such as anodizing treatment cannot effectively solve this problem. Therefore, it is urgent to develop a systematic approach to mitigate such corrosion attack.

2. Experimental procedures

A 7075-T6 aluminum alloy sheet of 2 mm thickness was used. FSW was carried out by using home-optimized process parameters (shaft shoulder diameter of 10 mm; pin length of 1.8 cm; rotational speed of 1200 rpm; welding speed of 44 mm min⁻¹). The sample with dimensions of 30′*15′*2 mm³ was intercepted by wire cutting with the weld at the central line of the sample. After preliminary grinding, the samples were sealed with epoxy resin and ready for further use.

Samples were ground sequentially with metallographic abrasive paper from 240 # to 3000 #. After mechanical polishing, the samples were etched with Keller reagent (a mixture of 1 ml HF, 1.5 ml HCl, 2.5 ml HNO₃ and
95 ml H₂O). After cleaning and drying, the microstructure of the weld was firstly examined by a LECAM5000 optical microscope (OM), with measurement of grain size by linear intercept method. For microstructure characterization by scanning electron microscopy (SEM), the surface and cross section of the weld were electropolished with AC₄₂ solution (10ml HClO₄ and 910 ml C₂H₅OH) at −30 °C, at a voltage of 20 V. Microhardness test was conducted on the cross-sectional plane with a spacing of 0.5 mm for each point along the central line of the plate thickness, in the horizontal direction, by using an HVS-1000 Vickers indenter with an applied load of 200 g for 10 s.

Corrosion immersion test was carried out in 3.5 wt% NaCl aqueous solution at room temperature for up to 9 h. After immersion, the samples were immersed in a 30 vol.% nitric acid solution for 30 s to remove the corrosion products. Then the corrosion morphology of the alloy was examined by OM and SEM. Anodizing was carried out on a KR50003−50V/3A DC power supply, at a constant voltage of 14 V, in a mixed solution containing 0.53 M tartarate and 0.46 M sulfuric acid, at 37 °C, for 1500 s. The sealing treatment of the anodic film was carried out in 0.01M lithium carbonate aqueous solution, at 50 °C, for 1800 s.

A GAMRY Reference 3000 potentiostat was used for potentiodynamic polarization (PDP) test. A saturated calomel electrode (SCE), a platinum sheet and the sample were used as reference electrode, counter electrode and working electrode, respectively. Before the PDP measurement, the sample was immersed in 3.5 wt%NaCl solution for 3600 s to obtain a relatively stable open circuit potential (OCP). The PDP was then performed by polarizing the sample from −0.1 to 2 V (OCP) at a scanning rate of 0.5 mV s⁻¹. The corrosion behavior of the alloy, after anodizing and sealing, in 3.5 wt% NaCl solution was also investigated by scanning vibrating electrode technology (SVET) on a Princeton VersaSCAN system. The microelectrode was located at 200 μm above the sample surface and vibrated at 80 Hz, with an amplitude of 30 μm and a step length of 100 μm. A Zeiss SIGMA HD field emission scanning electron microscope, equipped with energy dispersive spectroscopy (EDS), was used to examine the general microstructure of samples after different treatments. The cross sections of samples were prepared by ultramicrotomy (Leica Ultracut) using a diamond knife. In order to characterize the strengthening precipitates by transmission electron microscopy (TEM), thin plates of 100 μm extracted from the characteristic zones of the weld joint were twin-jet electropolished in a solution containing 30 vol.% nitric acid and 70 vol.% methanol at −30 °C.

3. Results and discussion

3.1. Microstructure of the FSW joint

Figure 1 shows optical micrographs of the FSW joint. The characteristic zones of the FSW joint, namely NZ, TMAZ and HAZ, were clearly revealed in figure 1(a). The NZ was in bowl-like shape and showed the typical ‘onion ring’ morphology [41]. Figures 1(b)−(c) show the optical micrographs of the BM, HAZ, TMAZ and NZ at higher magnifications, respectively. As shown in figure 1(b), the grains in the BM were elongated along the rolling direction, decorated with dark spots corresponding to FeCu₂Al₇ or Fe₃Al intermetallic particles (IMPs) formed during the solidification process [16]. The shape of the grains in the HAZ (figure 1(c)) was like that in the BM, however, the contrast difference of the grains in the HAZ suggests change of microstructure. The grains in the TMAZ were not distributed uniformly, with relatively finer grains in regions adjacent to the NZ and relatively coarser grains in the region adjacent to the HAZ (figure 1(d)). In the NZ, the grains were too small to be clearly resolved by the optical microscope (figure 1(e)).

Figure 2 compares typical backscattered electron micrographs of the BM, HAZ, TMAZ and NZ at high magnifications. As shown in figure 2(a), many spherical precipitates of 100 ~ 150 nm diameters were uniformly distributed within the grains, and the grain boundaries were relatively clean. The distribution of precipitates in the HAZ were like that in the BM except that grain boundary precipitates were clearly revealed in the HAZ (figure 2(b)). In TMAZ (figure 2(c)), in addition to fine spherical precipitates, more coarse grain boundary precipitates were observed. It was noticed that the coarse grain boundary precipitates were distributed heterogeneously, with a higher population density around the severely deformed grains (as indicated by the dashed line circle). This suggests that the plastic deformation in TMAZ is not uniform and the heat arising from friction and deformation has resulted in coarsening and/or dissolution of precipitates locally. Fine equiaxed grains of about 1 μm in diameters were revealed in the NZ, corresponding to dynamic recrystallization grains. Large number of spherical precipitates were also present in the NZ. According to the literature [42, 43], the spherical precipitates correspond to S (Al₃CuMg) phase while the grain boundary precipitates mainly correspond to η (MgZn₂) phase.

Figure 3 shows bright field TEM images taken from different zones of the FSW joint. According to the literature [44], the fine precipitates in the grain interior in BM were mainly η’ (MgZn₂) phase, and the fine and discontinuous precipitates at grain boundaries were η phase (MgZn₂) (figure 3(a)). In the HAZ (figure 3(b)) and TMAZ (figure 3(c)), the η’ phase in the grain interior and η phase at grain boundaries became coarser due to elevated temperature in these regions. Additionally, with the growth of grain boundary η phase, their population...
density decreased and the width of the precipitate free zones (PFZ) increased accordingly. In the NZ (figure 3(d)), due to the significant increase of temperature, η/’ phase dissolved into the alloy matrix while spherical S phase re-precipitated during the cooling process [45]. It should be noted that some individual coarse particles were frequently observed at grain boundaries in the HAZ and TMAZ, as indicated by the white arrows in figures 3(b) and (c). EDS mapping analysis suggested that these particles were rich in Cu and Zn (figure 4). It was reported that such grain boundary particles were related to diffusion of copper from the grain interior to the grain boundary precipitates at 200 °C–300 °C [46]. Probably, Cu might have substitute Mg in the grain boundary η phase, leading to formation of the new grain boundary precipitates that are rich in Cu and Zn.

Microstructure difference in each zone was also reflected by the hardness distribution. Typically, the hardness distribution was revealed as a ‘W’ shape across the FSW joint on the cross section (figure 5). The transitional zone between the HAZ and TMAZ had the lowest hardness (~125 HV_{0.2}) while the BM had highest hardness (~185 HV_{0.2}). The hardness of the NZ (~150 HV_{0.2}) was between that of BM and TMAZ/HAZ. According to figures 2 and 3, no plastic deformation occurred in HAZ but η/’ phase was coarsened and converted into stable phase, resulting in decreasing hardness in HAZ [47]. The severe plastic deformation of TMAZ under the action of...
of shear force makes the grains slender and solute diffusion segregation coarsen or dissolve due to the high temperature, which leads to significant hardness decline at the shoulder of TMAZ. The hardness of NZ is lower than that of BM because most of the strengthening phases such as $\eta'$/\(\eta\) have dissolved completely in NZ [48].

### 3.2. Corrosion behavior of the FSW joint

Figure 6 shows the corrosion morphologies of the FSW joint after immersing in 3.5 wt% NaCl solution for 9 h. As shown in figure 6(a), stable localized corrosion (SLC) occurred in the transition zone between the TMAZ and HAZ. Bright areas appeared around the localized corrosion sites, which was associated with the cathodic protection effect of the SLC to surrounding areas. The larger the cathodic protection area, the severer the localized corrosion. Thus, it is suggested that the localized corrosion was severer in the advancing side (AS) than the retreating side (RS). This is probably related to higher linear speed and consequently severer plastic deformation in the AS compared with the RS during welding [23]. The NZ became dark after the immersion test, suggesting uniform deposition of corrosion products (figure 6(b)). Occasionally, some small bright areas were revealed in NZ (figure 6(c)), just like the corrosion morphology in the BM (figure 6(d)). As suggested latter, these bright areas were caused by metastable localized corrosion (MLC) associated with coarse IMPs. This suggests that the plastic deformation was not uniform in the NZ and some large IMPs had not been broken into small ones during the welding process.

Figure 7 shows secondary electron micrographs of the FSW joint surface after immersion in 3.5% NaCl solution for 9 h. It is clearly indicated that coarse IMPs had caused MSL in the BM (figure 7(a)). The IMP-related MSL became less evident in the NZ due to the reduction in size and re-distribution of IMPs (figure 7(b)). Additionally, many spherical cavities of about 1 µm in diameters were observed in the NZ. Such spherical particles should be related to spherical S phase. The S phase was first corroded in a de-alloying process, leading to enrichment of copper at the site of the corroded S phase; then the enriched copper remanent would work as the cathode, accelerating anodic dissolution of the surrounding aluminum matrix. Figures 7(c) and (d) show typical SLC in the TMAZ/HAZ transition regions. The SLC were characterized by intergranular corrosion.

In order to understand the initiation and propagation of SLC, the cross sections of typical localized corrosion sites in the TMAZ/HAZ transition regions were prepared by ultramicrotomy and examined by SEM (figure 8). As suggested in figure 8(a), the localized corrosion was initiated on the alloy surface and propagated into the alloy matrix as deep as ~100 µm. Figures 8(b) and (c) show typical corrosion fronts at increased magnification, revealing evident grain boundaries attack. The surface (figures 7(c) and (d)) and cross-sectional morphology (figures 8(a)–(c)) of the localized corrosion sites jointly confirm that the SLC was closely related to grain boundary attack.
Figure 3. Bright field TEM images taken from different zones of the FSW joint: (a) BM; (b) HAZ; (c) TMAZ; (d) NZ.

Figure 4. EDS mapping analysis of individual coarse particles at grain boundaries in: (a) HAZ; (b) TMAZ.

Figure 8(d) shows backscattered electron micrograph of the framed area in figure 8(c), revealing coarse grain boundary precipitates at the corrosion fronts. Scrutiny of figure 8(d) indicated that the attached grain boundaries had a width of 500 ~ 1000 nm and individual bright particles were present in the middle of the corrosion path.
It is believed that the intergranular corrosion was related to the formation of Cu-rich grain boundary precipitates and the increasing width of the PFZ. The growth of the Cu-rich grain boundary precipitates required continuous supply of Cu and Zn from adjacent regions through a diffusion process. Consequently, the PFZ became wider and depleted in Cu and Zn. The Cu-rich grain boundary precipitates had much higher corrosion potential than the PFZs since the standard corrosion potentials of Cu\(/\text{Cu}^{2+}\) (+0.337 V) and Zn\(/\text{Zn}^{2+}\) (−0.763 V) are more positive than that of Al\(/\text{Al}^{3+}\) (−1.66 V). Thus, in corrosive medium, corrosion cells would be formed in grain boundary regions by the coupling effect between the Cu-rich grain boundary precipitates and the PFZs. The Cu-rich precipitates would work as cathode and accelerated the anodic dissolution of the PFZs, leading to intergranular corrosion.

Figure 9 shows the potentiodynamic polarization curves measured from different regions of the FSW joint in 3.5wt% NaCl solution. Since it is difficult to differentiate the HAZ and TMAZ during the test, these two zones were regarded as one region, i.e., TMAZ/HAZ. Table 1 shows the corrosion potential (E_{corr}) and corrosion...
current density ($I_{corr}$) of each region obtained by Tafel fitting. The BM exhibited the highest corrosion resistance, with the $I_{corr}$ of $3.289 \times 10^{-7}$ A cm$^{-2}$. The $I_{corr}$ of the NZ was about double of the BM, suggesting increased corrosion rate. This was probably ascribed to increased population density and size of the S phase in the NZ. The $I_{corr}$ of the TMAZ/HAZ was more than two orders higher than that of BM and NZ, suggesting significant increase of corrosion susceptibility in this region. The intergranular corrosion and the associated stable localized corrosion were responsible for the high corrosion current density.
3.3. Anodizing of the FSW joint

Figures 10(a)–(c) show the surface morphologies of FSW joints after TSA anodizing treatment. Micron-sized cavities were frequently observed on the sample surface (figures 10(a) and (b)), which were ascribed to anodizing and dissolution of the coarse IMPs. At increased magnification, typical nanopores in the anodic film and fine cavities of various shape were observed (figure 10(c)). According to microstructure characterization in figures 2 and 3, the spherical and rod-like cavities were related to anodized S phase and η (MgZn2) phase. Figure 10(d) shows the cross section of the anodic film at a typical cavity site, revealing a film thickness of 4.8 μm at the bottom of the cavity. Figures 10(e)–(h) shows typical cross sections of the anodic film formed in BM, HAZ, TMAZ and NZ. It was found that the anodic film was slightly thicker in the NZ (5.4 μm) than in the BM, HAZ and TMAZ (5.2 μm). This was probably because that most coarse IMPs had been broken into fine ones. As shown in figure 10(d), the anodizing and dissolution of the IMPs would result in reduced film thickness. The anodizing and dissolution of the fine IMPs in the NZ would have less effect on the film thickness, leading to larger film thickness and likely less cavity defects. Figures 11(a) and (b) show typical surface morphology of the anodic film after sealing in the lithium carbonate solution, at low and increased magnification, respectively. It was found that a lamellar feature was uniformly formed on the anodic film surface after sealing treatment. According to the literature [38–40], such a feature corresponded to the presence of layered double hydroxides (LDH), which had been found to provide effective sealing to the porous anodic film.

The corrosion resistance of the FSW joint after anodizing treatment was verified by both the immersion and neutral salt spray tests. The changes of corrosion potential/current density in the micro-area of the sample surface were recorded by scanning vibration electrodes (SVET) intermittently during the immersion test. Figure 12 compares the SVET mappings of the anodized alloy sample recorded during immersion for up to 192 h. The TSA sample exhibited stable localized corrosion in the transition zone between HAZ and TMAZ after immersion for 6 h. The initialized localized corrosion propagated rapidly with prolonged immersion, as suggested by the increased current density at the localized corrosion sites. In contrast, the alloy sample after LDH sealing did not show any localized corrosion even after immersion for 192 h. Figure 13 shows macroscopic appearance of the LDH-sealed alloy sample after neutral salt spray test for 192 h. No evident localized corrosion was found in the BM, suggesting that the anodizing treatment could provide enough protection to the 7075-T6 aluminum alloy. However, one SLC pit was found in the TMAZ/HAZ transition regions in the front and back side of the sample (as indicated by the circles and arrows). It was noted that no obvious accumulation of corrosion products presenting near the corrosion pits, suggesting that the corrosion attack was weak at this moment.

Table 1. Fitting data of polarization curves of samples of figure 8.

| Samples     | BM            | NZ            | TMAZ/HAZ         |
|-------------|---------------|---------------|-----------------|
| Icorr (A cm⁻²) | 3.289 × 10⁻⁷ | 6.023 × 10⁻⁷ | 1.036 × 10⁻⁵   |
| Ec (V)       | −0.754        | −0.834        | −0.877          |
As shown in figure 10, the thickness of the anodic film in the HAZ and TMAZ was comparable to that of the BM. Therefore, the relatively high corrosion susceptibility of the TMAZ/HAZ relative to the BM (figure 13) could not be ascribed to film thickness difference. On the other hand, since the major film defects were arising from anodized IMPs (figures 10(a)–(d)), the size of the cavity defects in the HAZ and TMAZ should be similar or even smaller than that in the BM. Therefore, the presence of film defects could not explain the higher corrosion susceptibility of the TMAZ/HAZ regions. Thus, it was concluded that the appearance of the localized corrosion sites in TMAZ/HAZ transition regions after the anodizing treatment should not be ascribed to structure of the anodic film. Instead, it is believed that the high corrosion susceptibility of the alloy substrate in the TMAZ/HAZ transition regions was responsible for the phenomenon. In other words, apart from surface treatment, the key

**Figure 10.** SEM micrographs showing the (a)–(c) surface and (d)–(h) cross-sectional morphology of the anodized FSW joint: (d) cavity defect in BM; (e) BM; (f) HAZ; (g) TMAZ; (h) NZ.
issue to improve the corrosion resistance of the FSW joint of 7075-T6 aluminum alloy is to optimize the micro-
structure of the TMAZ/HAZ transition regions. Therefore, it is necessary to pay continuous attention to the
localized corrosion issue in the FSW joint and to seek for new corrosion mitigation methods to solve this
problem.

Figure 11. Surface morphology of anodized film after LDH sealing.

Figure 12. SVET maps of the as-anodized (TSA) alloy and the alloy after anodizing and post-sealing (TSA + LDH) recorded during immersion in 3.5% NaCl solution for up to 192 h. The dashed lines indicate the position of artificial defects.

Figure 13. Macroscopic morphology of alloy sample (TSA + LHD) after 192 h salt spray test: (a) frontside of joint; (b) backside of joint.
4. Conclusions

(1) Three characteristic zones, namely NZ, TMAZ and HAZ were observed in the FSW joint of 7075-T6 aluminum alloy. The base metal mainly contained η′ and S phases in grain interior, and η phase at grain boundaries. The η′ and η phase coarsened in the HAZ and TMAZ. The growth of the grain boundary η phase and enrichment of copper in them resulted in widening of the PFZs and possibly elemental depletion in the PFZs. The grain boundaries were free of precipitates in the NZ and the S phase was found to be the main precipitates in grain interiors.

(2) Localized corrosion in the BM and NZ were metastable, which was related to Fe-containing IMPs and S phase. Stable localized corrosion occurred in the transition regions between the TMAZ and HAZ, which was typified by intergranular corrosion. The intergranular corrosion at the stable localized corrosion sites was ascribed to the galvanic coupling effect between Cu-rich grain boundary precipitates and the PFZs.

(3) After anodizing treatment, localized corrosion pits still appeared in the transition regions between the TMAZ and HAZ after exposure to the neutral salt spray test for 192 h. The high corrosion susceptibility of the FSW joint after anodizing treatment was not ascribed to the difference of the anodic oxide film but the heterogeneous microstructure of the alloy beneath the anodic film.

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Data availability statement

All data that support the findings of this study are included within the article (and any supplementary files).

ORCID iDs

Kai Li  https://orcid.org/0000-0001-8250-1770
Huabin Xu  https://orcid.org/0000-0002-1068-8473

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