

Pre-corrosion induced fatigue damage localization and failure in 2195-T8 Al-Li alloy laser-welded joints

Xinzhi Yang

PLA Rocket Force University of Engineering

Gan Tian

PLA Rocket Force University of Engineering

Dejun Liu

liudjwx801@yeah.net

PLA Rocket Force University of Engineering

Yongjie Cheng

PLA Rocket Force University of Engineering

Hongsheng Liu

PLA Rocket Force University of Engineering

Biyun Ren

PLA Rocket Force University of Engineering

Article

Keywords: Laser welding, Pre-corrosion, Fatigue damage localization, Digital image correlation, Failure behavior

Posted Date: August 21st, 2026

DOI: <https://doi.org/10.21203/rs.3.rs-10447175/v1>

License:   This work is licensed under a Creative Commons Attribution 4.0 International License.

[Read Full License](#)

Additional Declarations: Competing interest reported. Dejun Liu serves as a member of the Young Editorial Board of Surface Science and Technology. He was not involved in the journal's review of, or decisions related to, this manuscript. The other authors have no competing interests to declare.

Pre-corrosion induced fatigue damage localization and failure in 2195-T8 Al-Li alloy laser-welded joints

Abstract: Propellant-derived acidic media can couple localized corrosion with fatigue damage, threatening the long-term integrity of welded Al-Li alloy liquid-propellant tanks. Here, 2195-T8 Al-Li alloy laser-welded joints were pre-corroded in a simulated propellant-derived acidic environment. Region-specific electrochemistry, mechanical testing, in situ digital image correlation (DIC), and fractography were combined to resolve fatigue damage localization and failure paths. The welding thermal cycle dissolved or coarsened strengthening precipitates, promoted grain-boundary solute segregation, and formed coarse Cu-rich phases. Consequently, the fusion zone (FZ) and adjacent heat-affected zone (HAZ) became mechanically softened and electrochemically active. Relative to the base metal (BM), the FZ hardness decreased by approximately 37.5%, while its corrosion current density was 16.9 times the BM value. This electrochemically active region developed band-like strain localization during cyclic loading. Increasing pre-corrosion progressively interconnected isolated FZ damage into region-scale material loss, reducing joint strength, ductility, and fatigue life. DIC and fractography showed that pre-corrosion shifted crack initiation from the weld-softened zone to multiple corrosion defects. It also promoted microcrack coalescence along a coupled corrosion-softening weak band. The fraction of fatigue life between the DIC-detectable macroscopic crack and final fracture fell from 10.6% to approximately 5%. These results show that spatially coincident weld softening and electrochemical activation accelerate localized corrosion accumulation and fatigue damage, thereby governing the final failure path. The findings support damage assessment and life management of welded tank structures.

Keywords: Laser welding; Pre-corrosion; Fatigue damage localization; Digital image correlation; Failure behavior

1 Introduction

Liquid-propellant tank integrity during long-term storage and transportation is critical to launch-vehicle safety. Moisture uptake by dinitrogen tetroxide generates

acidic products such as nitric acid, while sustained vibration and static loads expose welds to localized corrosion and service-induced failure [1]. Third-generation Al-Cu-Li alloys combine low density with high specific strength and stiffness, making them attractive for load-bearing aerospace structures [2,3]. Laser welding efficiently joins complex aluminum-alloy components because it offers low heat input, a narrow heat-affected zone, and high automation [4–9]. However, weld-induced microstructural and property heterogeneity may interact with localized corrosion and cyclic deformation, causing weak joint regions to govern premature fatigue failure.

Rapid melting, solidification, and nonuniform thermal cycling during high-energy-beam welding can dissolve or coarsen nanoscale intragranular strengthening precipitates. These processes also cause Li/Mg burn-off, Cu segregation, and coarse grain-boundary second phases in Al-Cu-Li alloys [7,10–13]. Fusion-welded Al-Cu-Li joints consequently exhibit pronounced microstructural, hardness, and strength gradients, with tensile fracture typically occurring near the fusion line [12,14–17]. Although previous studies have established the metallurgical origins of joint softening and grain-boundary embrittlement, they have focused mainly on weld formation and static properties. Whether the softened zone is also preferentially corroded, and how the spatial overlap between softening and corrosion controls fatigue damage, remain unclear.

Jiang et al. [18] showed that microgalvanic corrosion associated with T_1 precipitates can promote adjacent-matrix attack through selective Li dissolution and phase-polarity reversal. Yang et al. [19] found that grain-boundary Cu segregation and Cu-rich T_B and θ phases in the FZ of laser-welded 2195-T8 joints induce preferential intergranular dissolution and corrosion-channel extension in acidic media. Fu et al. [20] reported that corrosion pits and associated embrittlement reduce the ductility and fatigue life of 2195 Al-Li alloy. Using three-dimensional surface topography, 3D-DIC, and fractography, Song et al. [21] linked corrosion micropits, grooves, and subsurface defects to multi-site crack initiation through local stress/strain concentration or embrittlement. In welded joints, Yadav et al. [22] attributed reduced fatigue life in friction-stir-welded 2024-T3 joints to localized HAZ corrosion. Wang et al. [23] further

showed that second-phase electrochemical interactions near the fusion line and stress concentrations at microstructural transitions synergistically promote crack initiation and growth in laser wire-filled 6082 joints. However, region-specific electrochemical activity has not been systematically linked to corrosion-damage interconnection, cyclic strain localization, and final fracture paths in laser-welded Al-Cu-Li joints.

Accordingly, this study examined the mechanical degradation of 2195-T8 laser-welded joints after pre-corrosion in a simulated propellant-derived acidic environment. TEM, microhardness testing, and region-specific electrochemical measurements established microstructural, mechanical, and electrochemical heterogeneity across the BM, HAZ, and FZ. Three-dimensional corrosion topography, tensile testing, and S–N analysis quantified corrosion-damage interconnection and losses in load-bearing capacity and fatigue life. Finally, in situ DIC and SEM fractography tracked strain localization, DIC-detectable macroscopic crack formation, crack growth, and final fracture after 24 and 48 h of pre-corrosion. This approach clarifies how selective corrosion amplifies the weld-softened zone and directs fatigue damage along a coupled corrosion-softening weak band.

2 Experimental procedures

2.1 Material and laser welding procedure

Table 1 lists the chemical composition of the 2195-T8 Al-Li alloy plate. Before welding, the plates were alkaline-cleaned to remove surface oil, rinsed with acetone, and air-dried. Welding used a KUKA KR 70 R2100 six-axis robot and a CWX-12000 fiber laser at 1800 W and $1.2 \text{ m} \cdot \text{min}^{-1}$. High-purity Ar shielded the weld face and root at 18 and $10 \text{ L} \cdot \text{min}^{-1}$, respectively. Figs. 1a and 1b show the welding arrangement and weld macrographs. The grain morphologies in Fig. 1d and Fig. 1e were obtained in previous study [19] from specimens in the same batch and produced using the same welding parameters.

Table 1 Chemical composition of 2195-T8 Al-Li alloy (wt.%)

Element (wt.%)	Cu	Li	Mg	Ag	Zr	Ti	Al
AA2195-T8	4.20	1.02	0.42	0.20	0.12	0.03	Bal.

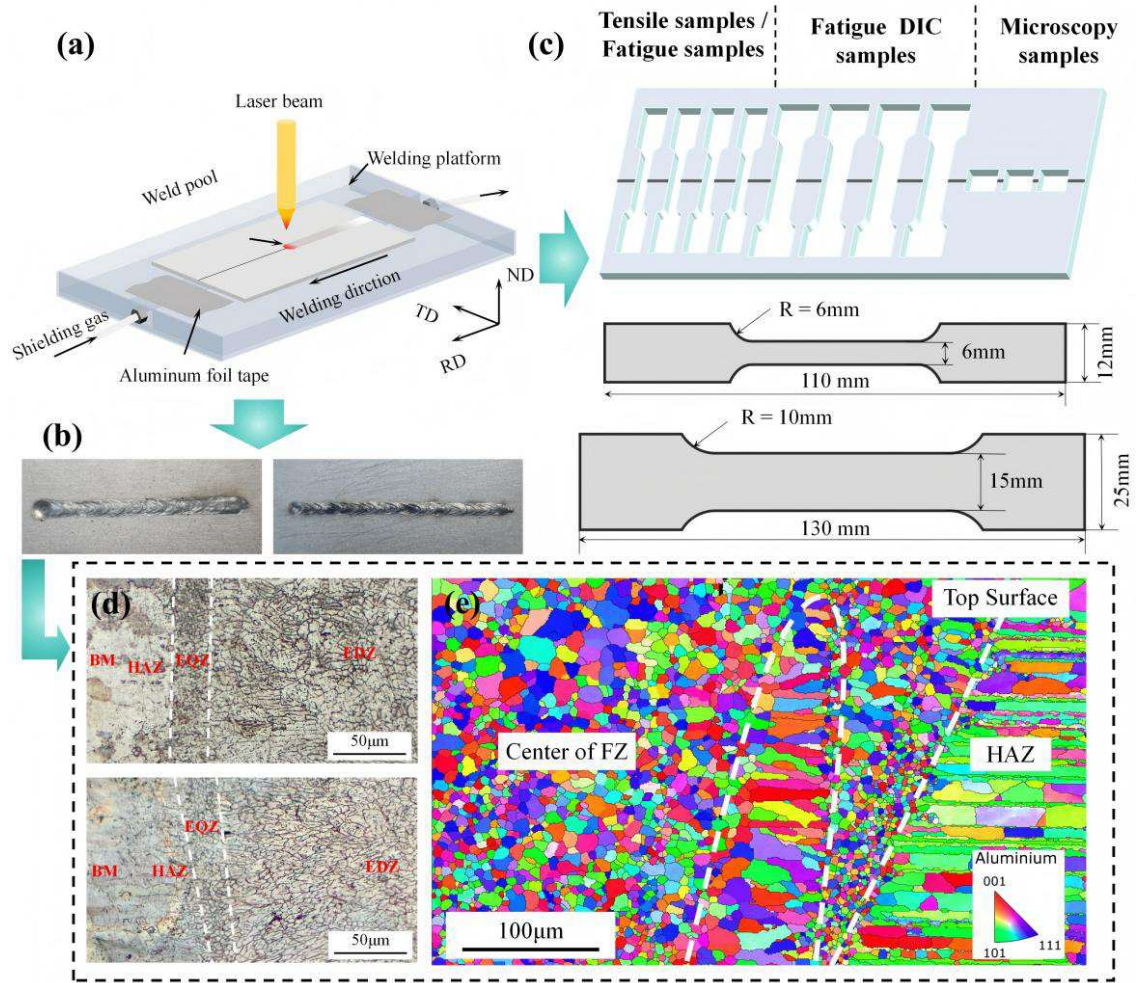


Fig. 1. Experimental workflow, specimen locations, and joint microstructural zones: (a) laser-welding setup; (b) weld-face and root macrographs; (c) sampling layout and dimensions of tensile, conventional fatigue, fatigue-DIC, and microstructural specimens; (d) top-surface and transverse joint microstructures; and (e) EBSD inverse pole figure (IPF) map from the FZ to the HAZ/BM.

2.2 Pre-corrosion treatment

Specimens retaining their as-welded surfaces were fully immersed in 3 wt.% HNO_3 at room temperature. Corrosion-morphology and tensile specimens were immersed for 4, 8, 24, 48, or 96 h; fatigue specimens were immersed for 24 or 48 h. After exposure, specimens were rinsed thoroughly with deionized water, and loosely adherent corrosion products were removed before testing.

2.3 Microstructural, electrochemical, and corrosion characterization

TEM specimens were extracted from the FZ, HAZ, and BM. Discs were mechanically ground to approximately 50 μm and ion-milled using a Gatan Model 695 precision ion-polishing system. Microstructure, phase constitution, and elemental

distributions were examined using an FEI Talos F200X field-emission TEM at 200 kV. Before microstructural and three-dimensional corrosion-topography measurements, specimens were ground sequentially with SiC paper to 2000 grit and polished with a 0.5 μm diamond suspension. They were then ultrasonically cleaned in acetone and cold-air dried. A confocal laser scanning microscope (CLSM; ZEISS LSM 900) recorded the joint microstructure and three-dimensional corrosion morphology. Vickers microhardness was measured along a line normal to the weld centerline at 0.2 mm intervals.

Electrochemical behavior of the polished BM, HAZ, and FZ was measured at room temperature in 3 wt.% HNO_3 using a three-electrode cell. The specimen, saturated calomel electrode (SCE), and platinum electrode served as the working, reference, and counter electrodes, respectively. Data were analyzed in NOVA 3.1, and potentiodynamic polarization was measured at $1 \text{ mV} \cdot \text{s}^{-1}$. Electrochemical impedance spectroscopy (EIS) covered 10^5 – 10^{-2} Hz, with 10 points per decade.

2.4 Mechanical testing and fractography

The sampling locations and specimen geometries are shown in Fig. 1c. Room-temperature tensile tests used a speed of $1 \text{ mm} \cdot \text{min}^{-1}$ and three specimens per corrosion condition. Constant-amplitude sinusoidal fatigue tests were conducted at 15 Hz and stress ratios $R = 0.1$ and 0.4, with three specimens per stress level. In situ DIC tracked fatigue damage in uncorroded specimens and those pre-corroded for 24 or 48 h. Tests were conducted at $R = 0.1$, an 80 MPa stress amplitude, and 15 Hz. DIC images were acquired while holding the peak load. After failure, fracture surfaces were examined by SEM without removing corrosion products, thereby preserving morphologies associated with corrosion-assisted crack initiation.

3 Results

3.1 Joint heterogeneity and region-selective corrosion morphology

Fig. 2 compares precipitates and grain-boundary chemistry in the BM, HAZ, and FZ to identify the microstructural origin of joint heterogeneity. The BM retained the

precipitation-strengthened structure typical of 2195-T8 Al-Li alloy, including lath-shaped T_i and fine, dispersed β'/δ' -type intragranular precipitates, as shown in Fig. 2a. Coarse intermetallic compounds were enriched mainly in Cu and Mg, as shown in Fig. 2b–h. The HAZ contained markedly fewer intragranular strengthening precipitates than the BM. Needle-like grain-boundary precipitates, Cu and Mg segregation, and isolated coarse intermetallic compounds were also observed, as shown in Fig. 2i–m. Thus, the welding thermal cycle altered both precipitation and grain-boundary chemistry. After melting and rapid solidification, the FZ contained fine dispersed phases and intragranular segregation bands, with continuous grain-boundary segregation and coarse Cu-rich intermetallic compounds, as shown in Fig. 2o–s. HREM indexing and our previous study [19] identified the coarse FZ grain-boundary phases as mainly T_B and θ . Overall, the joint transitioned from intragranular precipitation strengthening in the BM to precipitate depletion, grain-boundary precipitation, solute segregation, and coarse Cu-rich phases in the HAZ and FZ, as shown schematically in Fig. 3a.

Fig. 3b shows the microhardness profile associated with this heterogeneity. Consistent with the TEM observations, the precipitate-rich BM maintained a hardness of approximately 120 HV. Hardness fell sharply across the HAZ, while most of the FZ measured approximately 70–85 HV, with a minimum near the weld centerline. The retained T_i and β'/δ' -type precipitates preserved precipitation strengthening in the BM. By contrast, precipitate dissolution and coarsening, HAZ grain-boundary reprecipitation, and loss of the original FZ strengthening structure created a continuous softened zone across the FZ and adjacent HAZ.

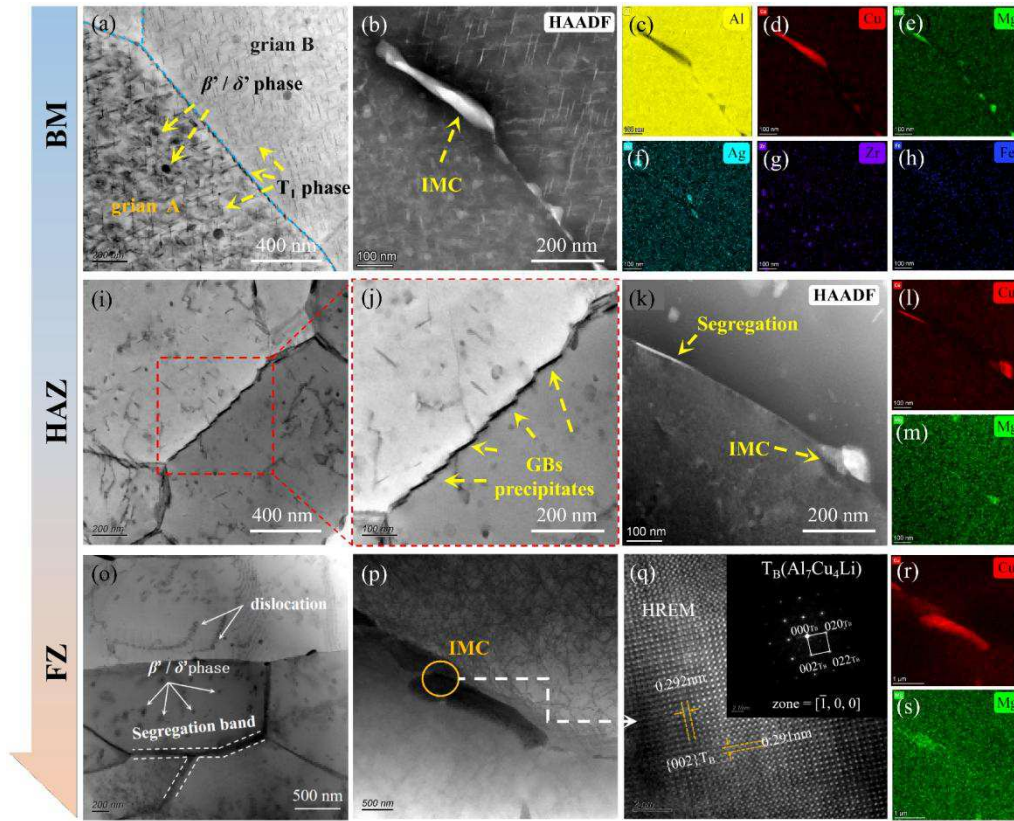


Fig. 2. TEM characterization of the BM, HAZ, and FZ in the 2195-T8 Al-Li alloy laser-welded joint: (a) BM grain-boundary region; (b) HAADF-STEM image of a coarse BM intermetallic; (c-h) corresponding Al, Cu, Mg, Ag, Zr, and Fe maps; (i) low-magnification HAADF image; (j) enlarged boxed region in (i); (k) HAADF-STEM image of an HAZ grain boundary; (l,m) corresponding Cu and Mg maps; (n) bright-field TEM image of an FZ grain boundary; (o) Cu-rich grain-boundary precipitate; (p) HREM image and diffraction-spot indexing of the marked region in (p); and (r,s) corresponding Cu and Mg maps.

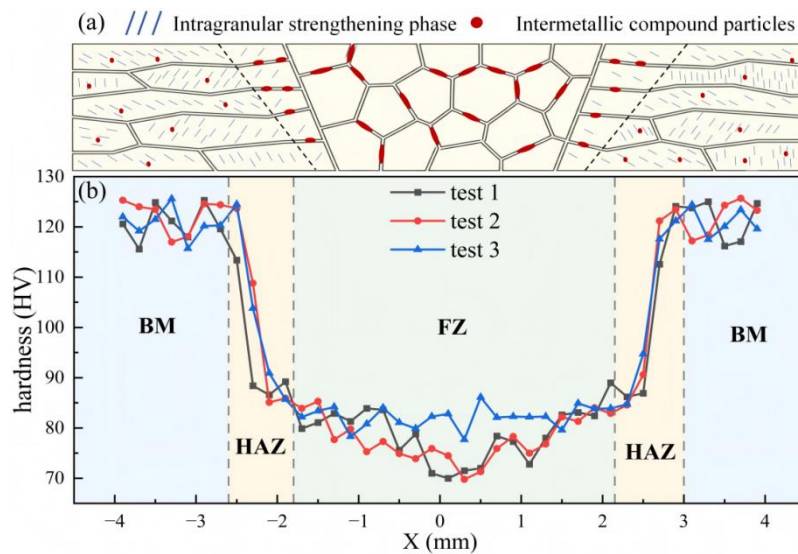


Fig. 3. Microstructural heterogeneity and hardness across the 2195-T8 Al-Li alloy laser-welded joint: (a) schematic BM, HAZ, and FZ microstructures; and (b) microhardness profiles.

Fig. 4 compares potentiodynamic polarization and EIS responses of the BM, HAZ, and FZ in 3 wt.% HNO₃. The corrosion potentials were approximately −0.289, 0.090, and 0.100 V vs. SCE for the BM, FZ, and HAZ, respectively. The FZ and HAZ corrosion current densities were 2.83×10^{-4} and 1.88×10^{-4} A·cm^{−2}, respectively. These values were 16.9 and 11.2 times the BM value (1.67×10^{-5} A·cm^{−2}), confirming greater electrochemical activity in the welded regions, particularly the FZ. Nyquist plots and fitted parameters were consistent with the polarization results. The BM, HAZ, and FZ had charge-transfer resistances (R_{ct}) of 2754, 844, and 209 Ω ·cm², respectively. The BM R_{ct} was therefore 13.2 times the FZ value and 3.3 times the HAZ value. Higher $CPE-Q$ and lower $CPE-n$ values in the FZ and HAZ further indicated more heterogeneous interfacial responses. Together, the corrosion current densities and charge-transfer resistances gave the following resistance order: BM > HAZ > FZ.

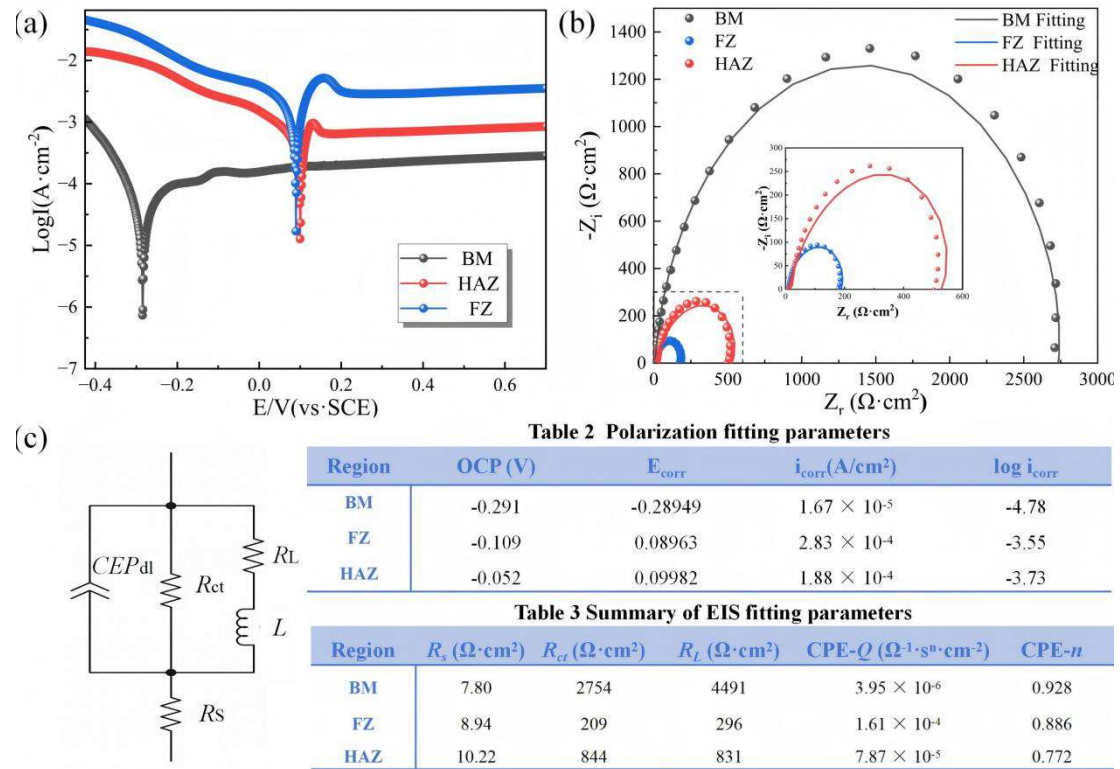


Fig. 4. Electrochemical behavior of the BM, HAZ, and FZ in the 2195-T8 Al-Li alloy laser-welded joint in 3 wt.% HNO₃: (a) potentiodynamic polarization curves; (b) Nyquist plots and fits; and (c) EIS equivalent circuit. Tables 2 and 3 list the fitted polarization and EIS parameters, respectively.

Fig. 5 shows the evolution of surface corrosion after 4–96 h in 3 wt.% HNO₃. After 4 h, fine corrosion traces followed the FZ solidification structure, whereas attack in the HAZ and BM remained limited. At 8 h, attack along FZ dendritic or cellular boundaries

intensified, and localized grain-boundary corrosion appeared in the HAZ. By 24 h, a more continuous dark network covered the FZ, and HAZ grain-boundary contours became clearer. The initially isolated damage had therefore interconnected along grain boundaries or solidification substructure boundaries. After 48 and 96 h, the FZ was extensively darkened, while the HAZ exhibited intergranular corrosion and pitting. Corrosion was therefore strongly region-selective and evolved from isolated damage into an interconnected network with increasing immersion time.

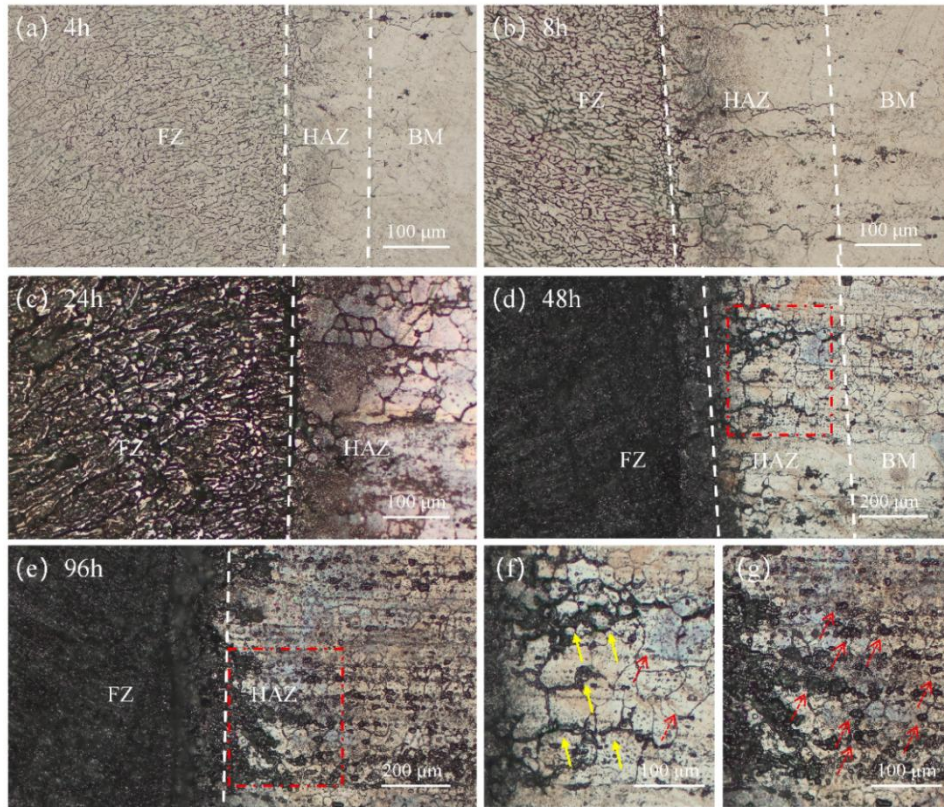


Fig. 5. Surface corrosion of the 2195-T8 Al-Li alloy laser-welded joint after 4–96 h in 3 wt.% HNO_3 : (a–e) 4, 8, 24, 48, and 96 h; and (f,g) enlarged red-boxed regions in (d,e). Yellow and red arrows mark HAZ intergranular corrosion and pitting, respectively.

Fig. 6 quantifies local material loss using three-dimensional topography and cross-sectional profiles. After 24 h, the FZ developed a clear height offset from adjacent regions. A representative profile gave a $10.97 \mu\text{m}$ height difference and a $16.45 \mu\text{m}$ transition width. The representative step height increased to $75.81 \mu\text{m}$ after 48 h and $191.32 \mu\text{m}$ after 96 h. Thus, prolonged pre-corrosion transformed surface topographic contrast into pronounced region-scale material loss.

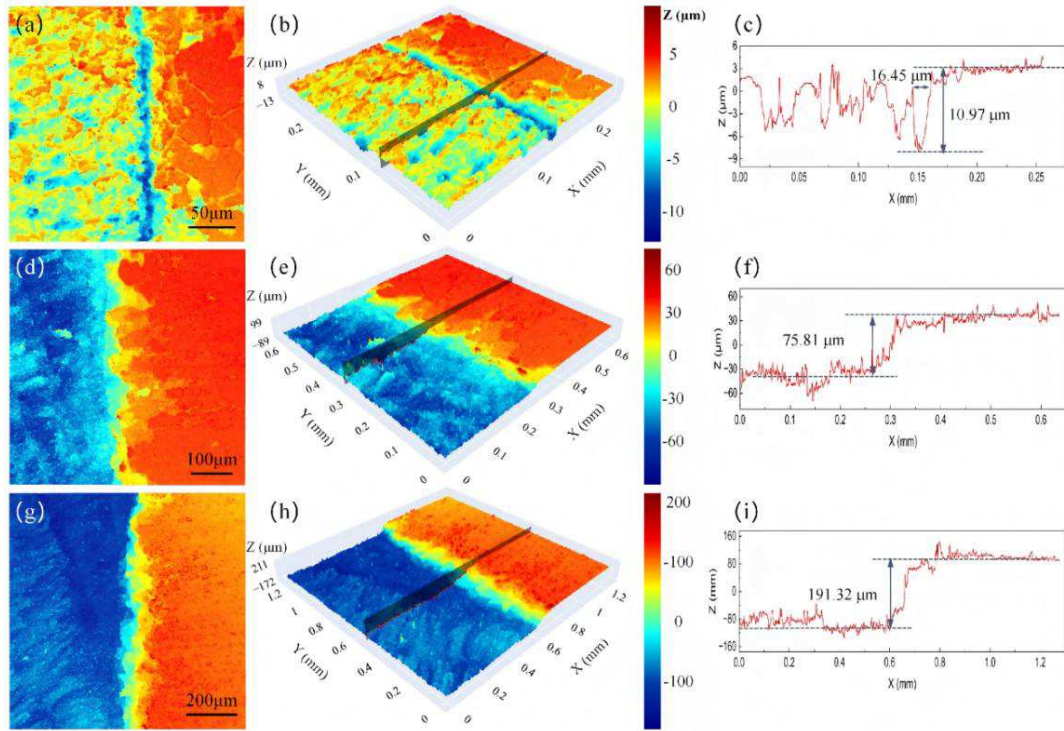


Fig. 6. Three-dimensional corrosion topographies and cross-sectional height profiles of the 2195-T8 Al-Li alloy laser-welded joint after 24, 48, and 96 h in 3 wt.% HNO₃: (a,d,g) two-dimensional height maps; (b,e,h) three-dimensional topographies; and (c,f,i) profiles along the marked paths.

3.2 Mechanical-property degradation after pre-corrosion

Fig. 7 summarizes the tensile behavior of joints under each pre-corrosion condition. Increasing exposure from 4 to 96 h reduced both peak stress and fracture strain, indicating concurrent losses in load-bearing capacity and ductility. The 96 h specimen fractured unstably at a relatively low stress and showed the greatest degradation. Yield strength, ultimate tensile strength, elastic modulus, and elongation after fracture all decreased with exposure time, as shown in Fig. 7b,c. Elongation showed the largest relative loss, indicating that joint ductility was most sensitive to localized corrosion.

Fig. 8 presents the joint stress-amplitude–fatigue-life relationships at $R = 0.1$ and 0.4 . Fatigue life increased as stress amplitude decreased at both ratios. At both stress ratios, pre-corrosion shifted the S–N datasets toward shorter lives. At comparable stress amplitudes, life generally decreased in the order uncorroded > 24 h > 48 h, confirming lower fatigue resistance after pre-corrosion.

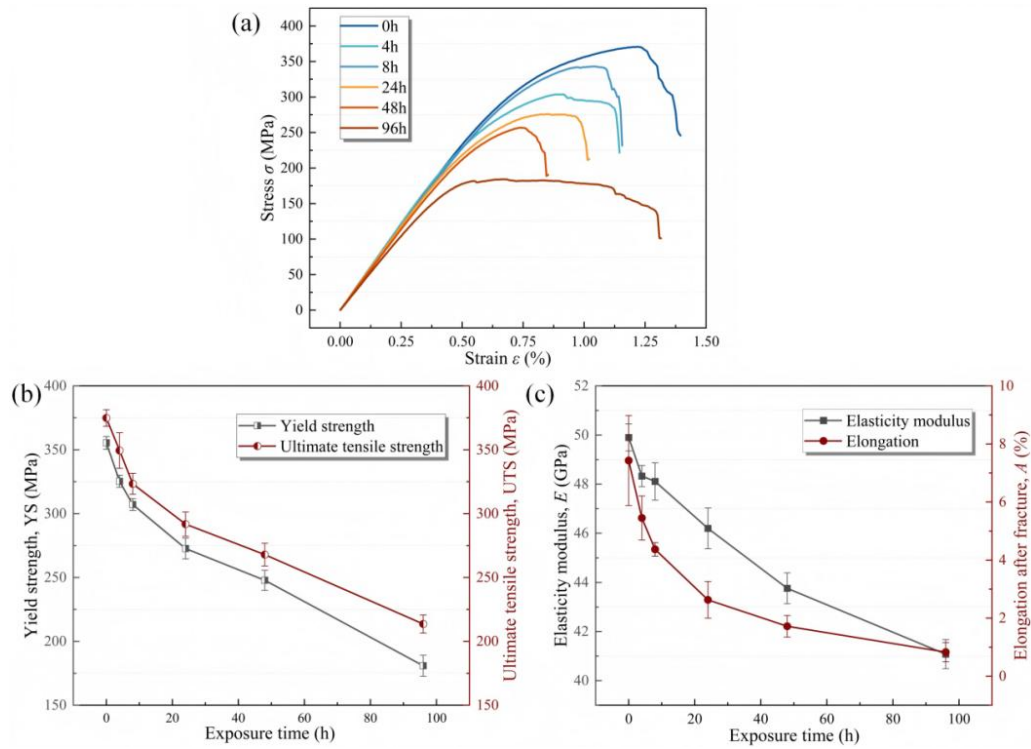


Fig. 7. Tensile properties of uncorroded and pre-corroded 2195-T8 Al-Li alloy laser-welded joints: (a) engineering stress–strain curves; (b) yield and ultimate tensile strengths; and (c) elastic modulus and elongation after fracture. Pre-corrosion times were 4, 8, 24, 48, and 96 h.

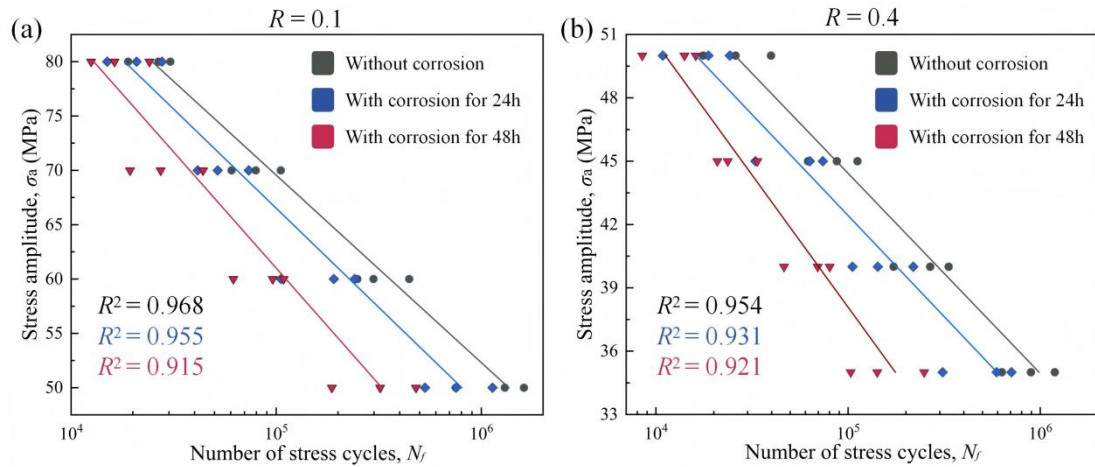


Fig. 8. Stress-amplitude–fatigue-life (S–N) relationships of uncorroded and pre-corroded 2195-T8 Al-Li alloy laser-welded joints: (a) $R = 0.1$; and (b) $R = 0.4$. Pre-corrosion times were 24 and 48 h.

Fig. 9 fits the fatigue-life data under each corrosion condition to a two-parameter Weibull distribution. The scale parameter η is the characteristic life at 63.2% failure probability, while β describes each dataset's distribution. At both stress ratios, 48 h pre-corrosion reduced η relative to the uncorroded condition, shifting the fitted distributions toward shorter lives. At $R = 0.4$, η fell by approximately 76%, from 2.87×10^5 to 6.91×10^4 cycles. Because each fit pooled data from different stress amplitudes,

β characterizes only the present dataset and was not extrapolated to intrinsic fatigue-life scatter.

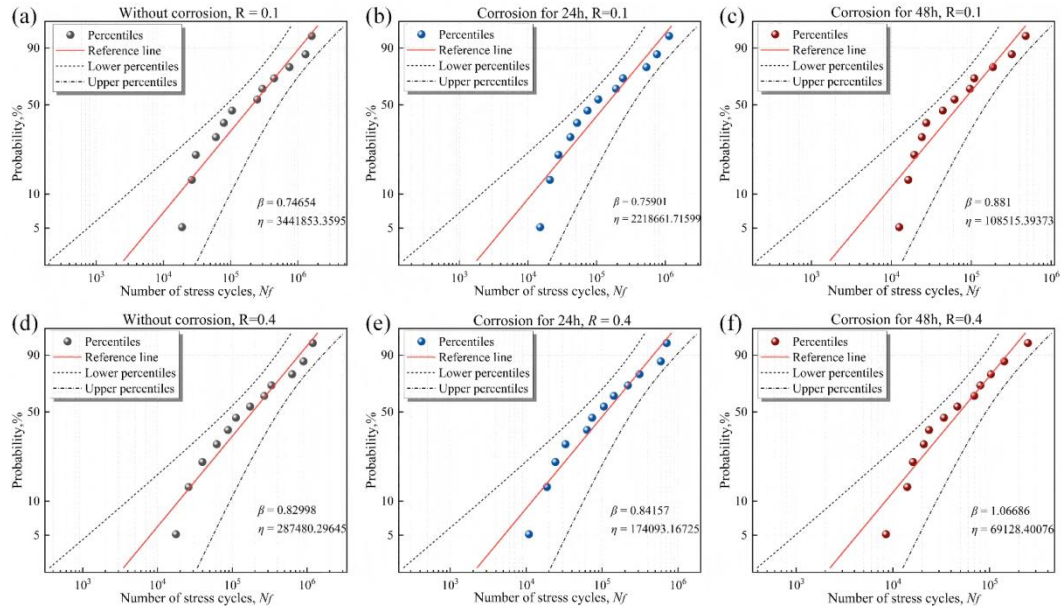


Fig. 9. Two-parameter Weibull fatigue-life distributions for 2195-T8 Al-Li alloy laser-welded joints: (a–c) uncorroded, 24 h, and 48 h pre-corrosion at $R = 0.1$; and (d–f) corresponding conditions at $R = 0.4$.

3.3 Evolution of cyclic strain localization

To resolve the spatial and stage-dependent evolution of fatigue damage, in situ DIC was performed at $R = 0.1$ and an 80 MPa stress amplitude, as shown in Figs. 10–12. Tests included uncorroded specimens and specimens pre-corroded for 24 or 48 h. S1–S5 denote a selected loading state, first DIC-detectable macroscopic crack, crack propagation, immediate pre-fracture state, and final fracture, respectively. S5 was used only to mark the final fracture location and cycle. All three uncorroded specimens developed a transverse band of localized axial strain across the weld and adjacent regions, as shown in Fig. 10. At S1, this band retained finite width and had not converged onto a single crack path. With cycling, localization intensified, and the crack propagated along the pre-existing high-strain zone to final fracture. Fatigue lives ranged from 15,424 to 25,577 cycles, indicating appreciable specimen-to-specimen scatter.

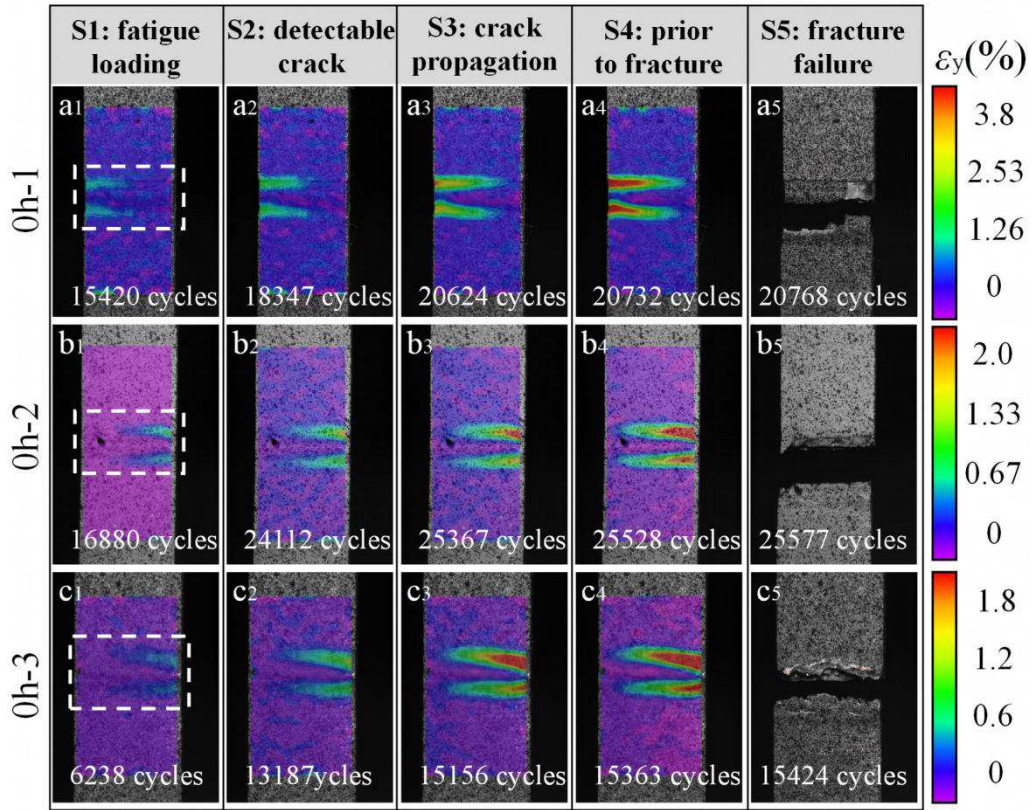


Fig. 10. Axial-strain evolution during fatigue cycling of three uncorroded 2195-T8 Al-Li alloy laser-welded joints at $R = 0.1$ and an 80 MPa stress amplitude.

After 24 h pre-corrosion, all three specimens developed transverse strain localization in the welded region, as shown in Fig. 11. In some specimens, the high-strain region progressively concentrated along a local weak band. Because S1 occurred at different absolute cycle counts and normalized life fractions, this trend was interpreted qualitatively. From S2 onward, crack-growth locations closely matched the preceding strain-localization zones and final fracture locations at S5. This correspondence indicates that corrosion pits and intergranular grooves formed during the 24 h exposure acted with the original weld-softened zone to select the macroscopic crack path.

After 48 h pre-corrosion, the high-strain regions intensified continuously along a fixed transverse weak band, as shown in Fig. 12. From S1 to S4, these regions followed essentially the same path and closely matched the final fracture locations at S5. Fatigue lives ranged from 10,580 to 17,246 cycles, lower overall than those of uncorroded specimens. Prolonged pre-corrosion interconnected isolated pits and intergranular grooves and caused region-scale material loss. This damage increasingly controlled the

macroscopic crack location and final fracture path.

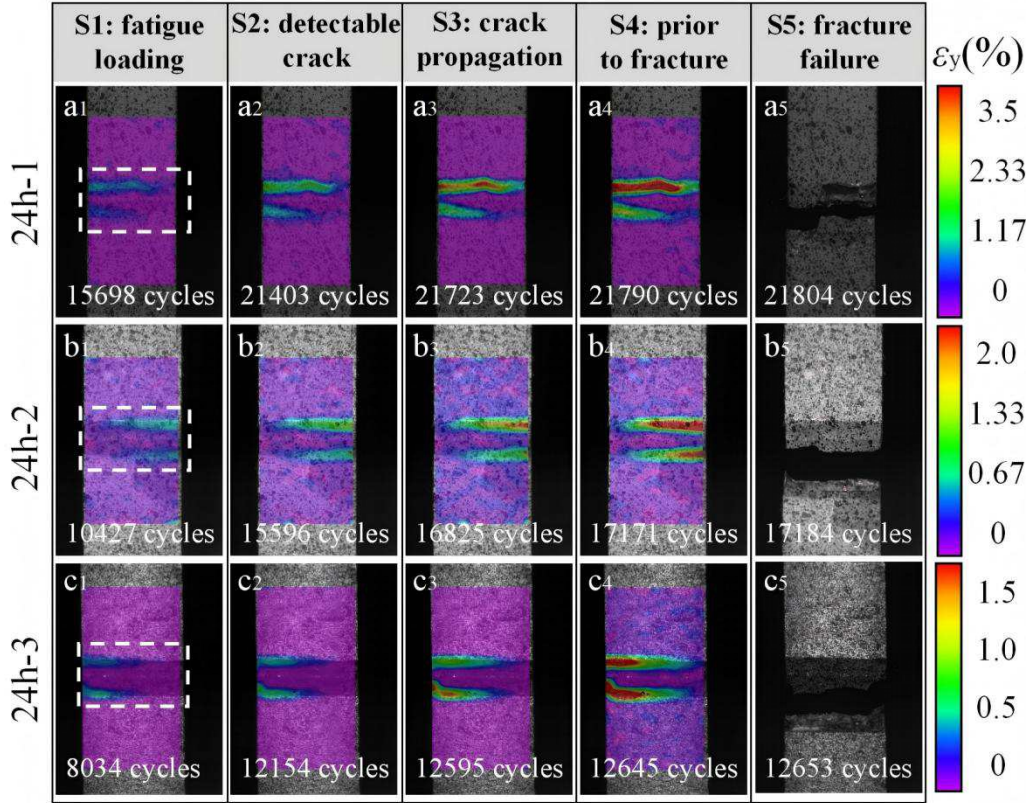


Fig. 11. Axial-strain evolution during fatigue cycling of three 2195-T8 Al-Li alloy laser-welded joints after 24 h pre-corrosion at $R = 0.1$ and an 80 MPa stress amplitude.

Fig. 13 compares the cyclic evolution of normalized mean strain in the welded region and the life fraction between the first DIC-detectable macroscopic crack and final fracture. Despite differing mean strain levels, all specimens showed a common trend. Regional mean strain remained relatively stable for most of the fatigue life, then rose sharply between crack detection and the pre-fracture state. This rise corresponded to crack growth and instability of the remaining ligament. S5 marked only the final fracture cycle and was excluded from quantitative strain comparisons. The S2–S5 interval averaged 10.60% of total life in uncorroded specimens, decreasing to 5.01% and 5.71% after 24 and 48 h pre-corrosion, respectively, as shown in Fig. 13d. Thus, pre-corrosion nearly halved the remaining-life window after DIC detected a macroscopic crack. Pre-existing microscale defects and reduced local load-bearing capacity therefore lowered joint residual damage tolerance. Whereas strain accumulated progressively within the softened zone in uncorroded joints, corrosion damage and weld softening jointly controlled localization and crack-path selection after

exposure.

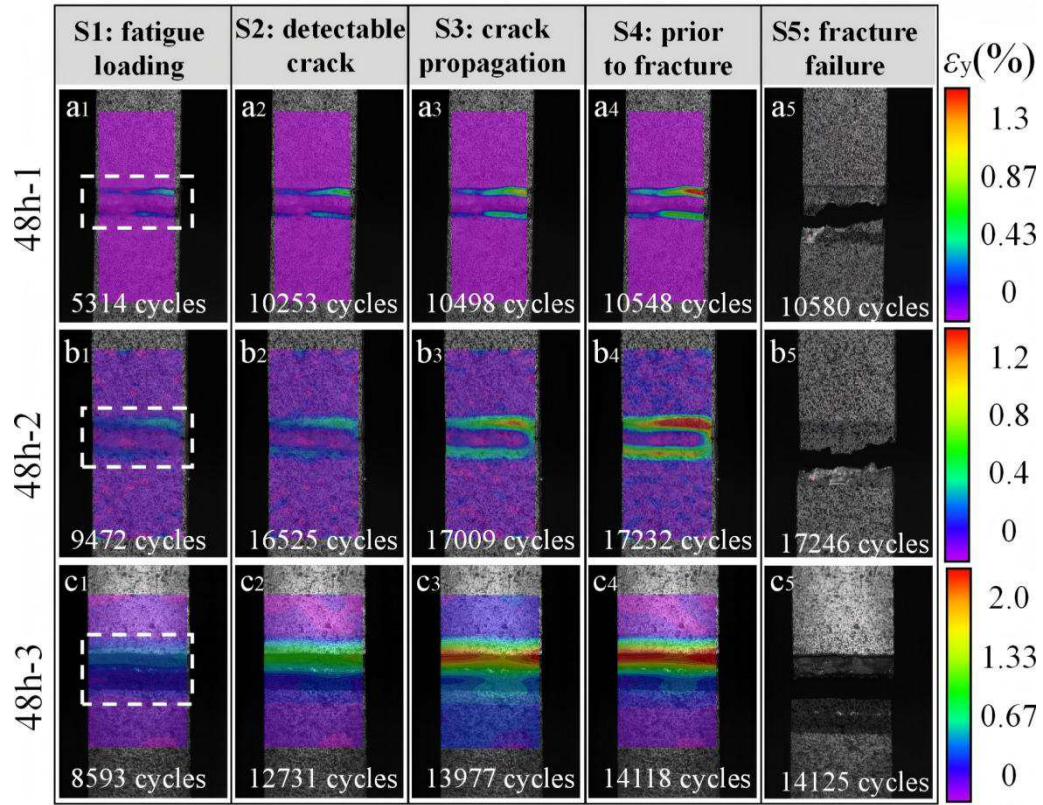


Fig. 12. Axial-strain evolution during fatigue cycling of three 2195-T8 Al-Li alloy laser-welded joints after 48 h pre-corrosion at $R = 0.1$ and an 80 MPa stress amplitude.

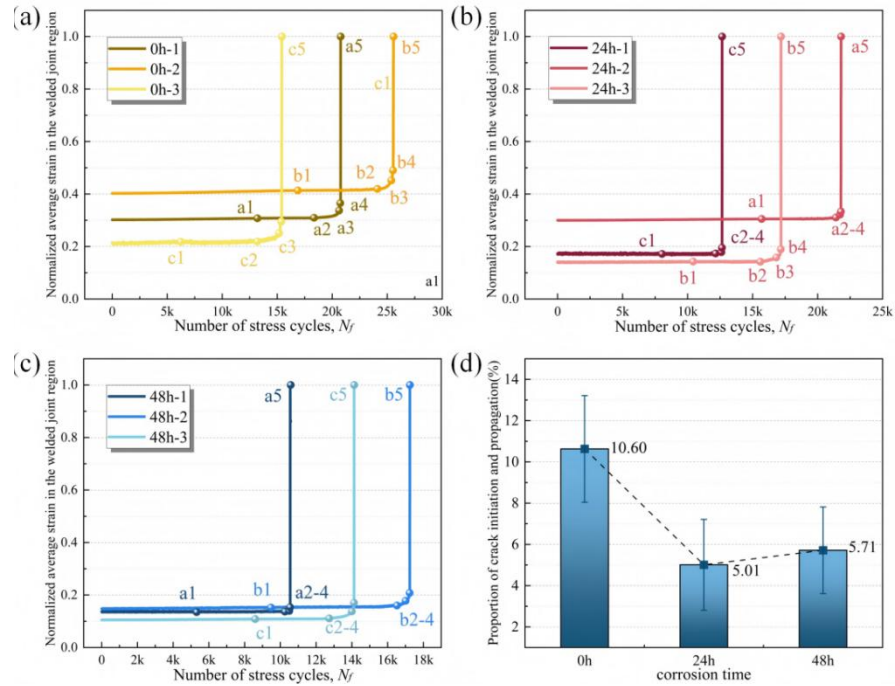


Fig. 13. Normalized mean strain in the welded region of 2195-T8 Al-Li alloy laser-welded joints during fatigue cycling: (a-c) uncorroded, 24 h, and 48 h pre-corrosion; and (d) life fraction from the first DIC-detectable macroscopic crack to final fracture. States a1-a5, b1-b5, and c1-c5 correspond to S1-S5 in Figs. 10-12.

3.4 Fractographic analysis

Fig. 14 shows the fatigue fracture surfaces of uncorroded joint specimens. Two representative low-magnification images show cracks initiating mainly at the top surface or near-surface regions before propagating inward. Each surface comprised crack-initiation, relatively flat stable-growth, and rough rapid-fracture regions. Wavy boundaries and step-like facets separated stable and rapid growth, indicating nonuniform crack-front advance caused by weld heterogeneity and local defects. Initiation regions contained serrated edges, microscale tearing ridges, and local pores or second-phase particles. Stable-growth regions showed clear fatigue striations, with representative spacings of approximately 0.86 and 1.14 μm in Figs. 14c and 14f, respectively, alongside fine dimples, secondary cracks, and step-like offsets. These features indicate progressive failure through initiation at mechanically weak surface sites, stable fatigue growth, and final rapid tearing.

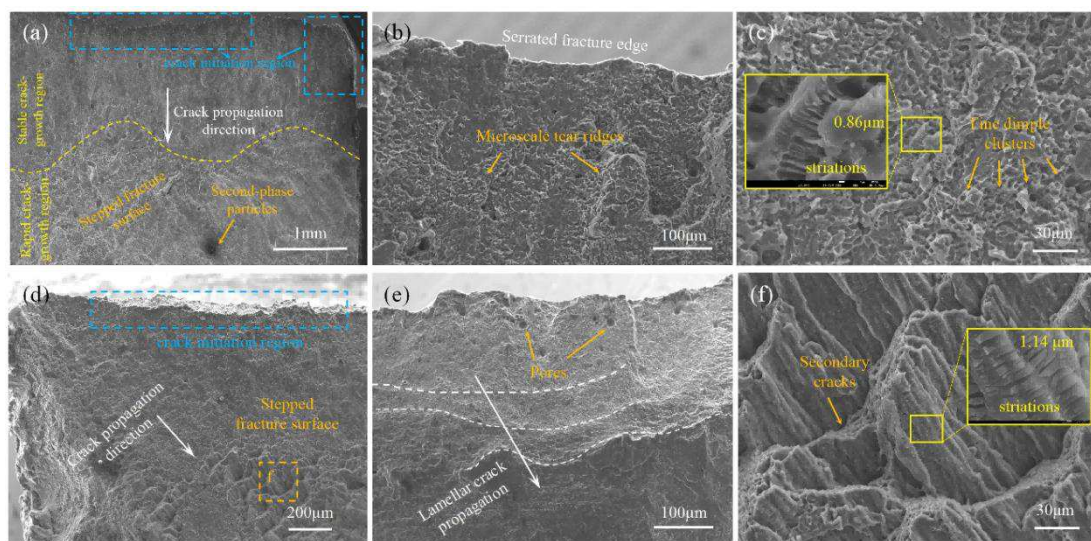


Fig. 14. Fatigue fracture surfaces of uncorroded 2195-T8 Al-Li alloy laser-welded DIC specimens tested at $R = 0.1$ and an 80 MPa stress amplitude: (a,d) low-magnification views of two representative specimens; (b,c) crack-initiation and stable-growth regions in (a); and (e,f) crack-initiation and step-like growth regions in (d).

After 24 h pre-corrosion, the fatigue fracture surface showed crack-initiation features associated with corrosion damage, as shown in Fig. 15. A continuous rough damage band formed near the top free surface. The crack grew inward from this band, deflected across a relatively flat fatigue-growth region, and entered the rapid-growth region. The initiation region contained clustered, nodular, loosely packed corrosion

products, an undulating morphology, and intergranular corrosion, as shown in Fig. 15b. Pre-corrosion therefore disrupted the surface layer, favoring initiation at pit bottoms and beside corrosion grooves. Step-like facets, secondary cracks, and second-phase particles indicate that multiple local crack fronts deflected, linked across offset planes, and coalesced, as shown in Fig. 15c. Fatigue striations and rapid-tearing morphology record continued growth after macroscopic crack formation, as shown in Fig. 15d. Compared with uncorroded specimens, 24 h pre-corrosion produced more pronounced crack coalescence, secondary cracking, and step-like propagation.

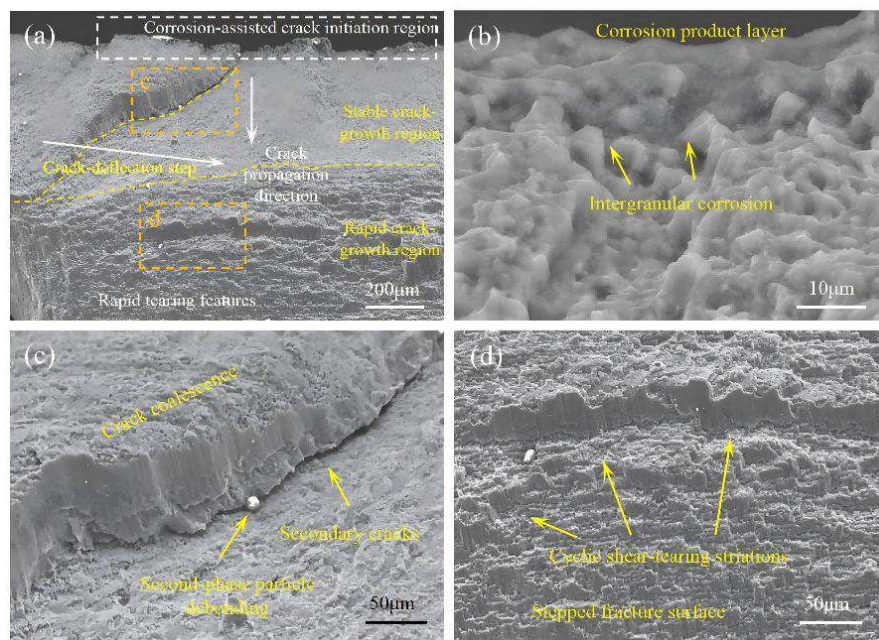


Fig. 15. Fatigue fracture surface of a 2195-T8 Al-Li alloy laser-welded DIC specimen after 24 h pre-corrosion, tested at $R = 0.1$ and an 80 MPa stress amplitude: (a) low-magnification view; and (b–d) corrosion-related initiation, crack-coalescence, and rapid-growth regions.

After 48 h pre-corrosion, localized corrosion damage occupied a larger fracture-surface area, as shown in Fig. 16. The surface comprised localized corrosion-damage, fatigue crack-growth, and rapid-tearing regions, with a clear transition between corrosion and mechanical fracture. Loose, spongy, clustered deposits, intergranular corrosion, and secondary cracks covered local areas, as shown in Fig. 16b. Pore-like defects also occurred at the base of this fracture-surface region. Prolonged pre-corrosion therefore enlarged the potential crack-source region, allowing multiple pits, grooves, and pores to contribute to early cracking. Across the transition region, corrosion products and rough porosity in the upper portion gave way to fine

dimples below, as shown in Fig. 16c. Once the crack penetrated the pre-corroded surface layer, mechanical fatigue and load transfer through the remaining ligament increasingly governed fracture. Elongated tearing ridges and a rough surface marked final rapid instability, as shown in Fig. 16d. Failure after 48 h pre-corrosion therefore involved multi-site initiation within extensive corrosion damage, followed by macroscopic crack growth and rapid tearing.

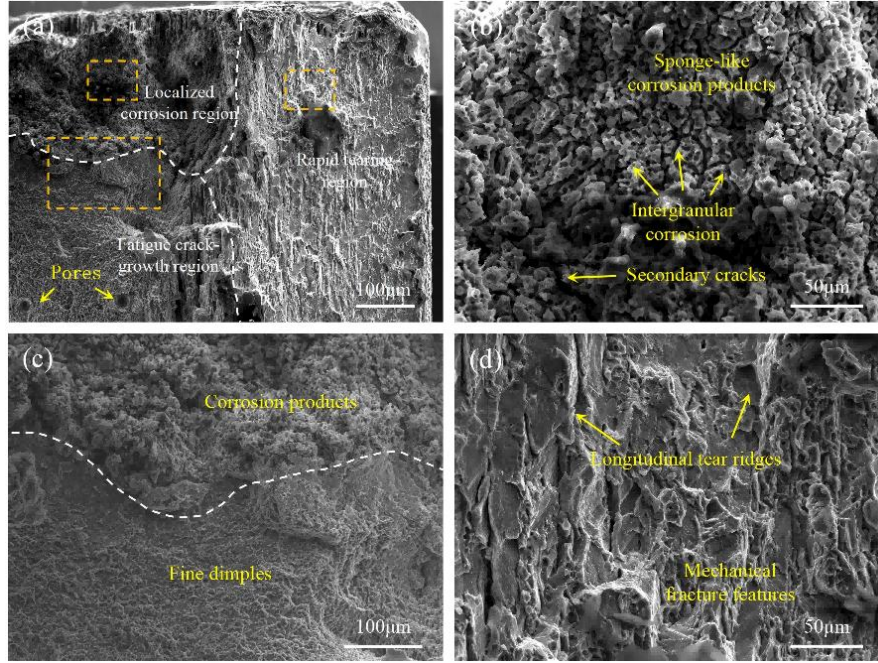


Fig. 16. Fatigue fracture surface of a 2195-T8 Al-Li alloy laser-welded DIC specimen after 48 h pre-corrosion, tested at $R = 0.1$ and an 80 MPa stress amplitude: (a) low-magnification view; and (b–d) localized corrosion-damage, corrosion-to-fatigue transition, and rapid-tearing regions.

4 Discussion

4.1 Formation and spatial coupling of microstructural, mechanical, and electrochemical weak zones

Laser welding generated pronounced gradients in grain morphology, precipitate and solute distributions, hardness, and electrochemical response [5,6,10,24]. Previous studies show that rapid solidification forms a fine equiaxed-grain band at the FZ edge with greater grain-boundary density than the weld center. Nonequilibrium solidification also segregates Cu along grain and substructure boundaries and forms coarse Cu-rich T_B and θ phases [11,19,25–28]. The high grain-boundary density supplies an interconnected network for intergranular corrosion. Segregation and electrochemical

contrast between Cu-rich phases and the matrix further activate these paths [18,26–32]. Meanwhile, T_1 dissolution or coarsening in the HAZ and loss of the original T8-strengthened FZ microstructure create a continuous softened zone [4,10,14,24,33]. In the present joint, hardness fell from approximately 120 HV in the BM to 70–85 HV in the welded region. The FZ and HAZ corrosion current densities were 16.9 and 11.2 times the BM value, respectively. Under cyclic loading, preferential deformation of the softened zone was constrained by the harder BM. This mismatch generated steep strain gradients and narrow high-strain bands near both fusion lines [34–36]. Preferential damage near the FZ edge and adjacent HAZ therefore cannot be attributed solely to minimum hardness or maximum electrochemical activity. Instead, it reflects the joint-scale overlap of high boundary density, solute segregation, mechanical softening, electrochemical activation, and strain localization.

4.2 Pre-corrosion amplification of the coupled weak zone and path-selective fatigue failure

Pre-corrosion began as discrete attack along FZ grain and solidification substructure boundaries, extended into the HAZ, and progressively interconnected [19,32,33]. Three-dimensional topography showed a relatively continuous FZ corrosion network after 24 h, with a representative height difference of 10.97 μm . At 48 h, this difference reached 75.81 μm , marking a transition from isolated grooves and pits to region-scale material loss. Corrosion grooves, surface steps, and cross-sectional loss disrupted local load transfer and strain compatibility, intensifying stress and strain concentrations. These changes accord with the measured losses in strength and ductility and the shift of the S–N curves toward shorter lives [37–41]. In uncorroded joints, DIC showed that a finite-width high-strain region within the softened zone gradually converged onto a macroscopic crack path. After 24 h pre-corrosion, pits and intergranular grooves began to select the crack location. After 48 h, the high-strain region followed essentially the same weak band from S1 to S4 and aligned more closely with final fracture [21,35,36,40,42,43]. Fractographic deposits, intergranular corrosion, secondary cracks, and step-like linkages corroborate this crack-source and growth sequence. Pre-corrosion therefore shifted initiation from the weld-softened zone to

multiple corrosion defects. It also promoted microcrack coalescence into a dominant crack along the coupled corrosion-softening band [1,37,40,42,43]. The life fraction from the first DIC-detectable macroscopic crack to fracture fell from 10.60% to approximately 5%, sharply narrowing the residual-life window.

5 Conclusions

This study examined how pre-corrosion affects mechanical properties, cyclic strain localization, and fatigue failure in 2195-T8 Al-Li alloy laser-welded joints. The principal conclusions are:

(1) Laser welding dissolved or coarsened strengthening precipitates, segregated solutes at grain boundaries, and formed coarse Cu-rich phases. These changes created a coupled microstructural, mechanical, and electrochemical weak zone across the FZ and adjacent HAZ. Relative to the BM, FZ hardness was approximately 37.5% lower, corrosion current density was 16.9 times the BM value, and charge-transfer resistance was only 7.6% of the BM value.

(2) In the simulated propellant-derived acidic environment, corrosion initiated preferentially along FZ grain and solidification substructure boundaries. Isolated intergranular attack and pits progressively interconnected into a continuous network and caused region-scale material loss. This evolution reduced joint strength and ductility and shifted the S–N curves toward shorter lives.

(3) DIC and fractography showed that pre-corrosion shifted crack initiation from the weld-softened zone to multiple corrosion defects. It also promoted microcrack coalescence and growth along the coupled weak band. The life fraction between the first DIC-detectable macroscopic crack and final fracture fell from 10.60% to approximately 5%. Thus, pre-corrosion reduced the joint's residual damage tolerance and increased the influence of the localized corrosion-damage zone on the final failure path.

References

- [1] B. Ren, G. Tian, D. Liu, X. Yang, M. Liu, Y. Zhang, Corrosion fatigue characteristics of 2195-T8 Al-Li alloys in an equivalent corrosive environment of propellant, *Engineering Failure Analysis* 180 (2025)

109916. <https://doi.org/10.1016/j.engfailanal.2025.109916>.

- [2] Y. Yang, J. Bi, H. Liu, Y. Li, M. Li, S. Ao, Z. Luo, Research progress on the microstructure and mechanical properties of friction stir welded Al–Li alloy joints, *Journal of Manufacturing Processes* 82 (2022) 230–244. <https://doi.org/10.1016/j.jmapro.2022.07.067>.
- [3] Y. Huang, P.-L. Liu, H. Xiang, S.-X. Deng, Y.-J. Guo, J.-F. Li, Mechanical properties, corrosion and microstructure distribution of a 2195-T8 Al–Li alloy TIG welded joint, *Journal of Manufacturing Processes* 90 (2023) 151–165. <https://doi.org/10.1016/j.jmapro.2023.02.007>.
- [4] D. Wu, C. Jia, D. Liu, X. Luo, F. Niu, G. Ma, Microstructure evolution and mechanical property of laser welded Al–Li alloy with filler strip, *Optics and Laser Technology* 165 (2023) 109585. <https://doi.org/10.1016/j.optlastec.2023.109585>.
- [5] Z. Luo, S. Zhao, L. Liu, H. Wang, X. Tan, X. Liang, H. Li, Optimization of microstructure and mechanical properties in laser welding of 2 mm-thick AA2195-T6 Al–Li alloy through defocusing control, *Journal of Materials Research and Technology* 40 (2026) 2437–2450. <https://doi.org/10.1016/j.jmrt.2025.12.276>.
- [6] C. Han, P. Jiang, S. Geng, L. Guo, K. Liu, Inhomogeneous microstructure distribution and its formation mechanism in deep penetration laser welding of medium-thick aluminum-lithium alloy plates, *Optics and Laser Technology* 167 (2023) 109783. <https://doi.org/10.1016/j.optlastec.2023.109783>.
- [7] C. Han, P. Jiang, S. Geng, L. Ren, J. Jin, Ultra grain refinement and mechanical properties improvement of all-weld-metal for medium-thick Al–Li alloy via laser beam oscillation and in-situ alloying, *Optics and Laser Technology* 168 (2024) 109965. <https://doi.org/10.1016/j.optlastec.2023.109965>.
- [8] F. Wu, Y. Zhu, B. Yuan, L. Li, In situ monitoring of stress corrosion at the Alloy 625|NaCl interface, *Surface Science and Technology* 2 (2024) 3. <https://doi.org/10.1007/s44251-024-00034-9>.
- [9] Q.-D. Zhong, X. Wang, Investigation of localized corrosion behavior in a copper-stainless steel curved surface coupling system in chloride-containing acidic environments using wire beam electrode techniques, *Surface Science and Technology* 3 (2025) 22. <https://doi.org/10.1007/s44251-025-00087-4>.
- [10] J. Bi, Z. Liu, J. Chi, H. Wang, C. Tan, X. Jia, Z. Yang, M.D. Starostenkov, G. Dong, Formation mechanisms and control strategies of FQZ softening in Al–Li alloy welded joint, *Journal of Materials Research and Technology* 23 (2023) 2810–2823. <https://doi.org/10.1016/j.jmrt.2023.01.147>.
- [11] Y. Shao, X. Li, L. Chen, E. He, New insights into fine equiaxed zone of laser-welded 2A97 Al–Li alloy, *Journal of Alloys and Compounds* 895 (2022) 162717. <https://doi.org/10.1016/j.jallcom.2021.162717>.
- [12] Y. Zhao, L. Hao, R. Liu, J. Wang, Y. Zeng, X. Zhan, Unravelling asymmetrical microstructure evolution and tensile fracture mechanism in laser welding of dissimilar 2219/2195 aluminum alloys, *Journal of Materials Research and Technology* 31 (2024) 4041–4058. <https://doi.org/10.1016/j.jmrt.2024.07.087>.
- [13] Z. Shao, M. Ma, X. Liu, et al., How do we improve the reproducibility and accuracy of EIS and Tafel testing in corrosion science? *Surface Science and Technology* 3 (2025) 3. <https://doi.org/10.1007/s44251-024-00066-1>.
- [14] G. Cui, C. Yang, X. Zhang, J. Bi, Microstructure evolution and mechanical properties of AA2195-T8 Al–Li alloy back gas-shielded VPPA weld, *Journal of Manufacturing Processes* 106 (2023) 288–302. <https://doi.org/10.1016/j.jmapro.2023.10.009>.
- [15] Y. Li, H. Bu, W. Ling, Y. Zhao, J. Wang, X. Zhan, Microstructure evolution and fracture behavior of 2060 Al–Li alloy by laser welding with TiC nanoparticle-reinforced filling wire, *Journal of Manufacturing Processes* 102 (2023) 67–78. <https://doi.org/10.1016/j.jmapro.2023.07.012>.
- [16] Y. Zhao, Y. Li, L. Ma, J. Wang, L. Hao, L. Wang, X. Zhan, Microstructure evolution and crack prevention in TiC nanoparticles-enhanced 2195 Al–Li alloy joint fabricated by coaxial fiber-diode hybrid laser wire-feeding welding, *Journal of Manufacturing Processes* 124 (2024) 566–580. <https://doi.org/10.1016/j.jmapro.2024.06.026>.
- [17] X. Zhan, Y. Li, Y. Liu, J. Wang, Z. Xu, Y. Zhao, X. Li, Controlling the columnar-to-equiaxed transition and crack propagation behavior of laser welded Al–Li alloy reinforced with TiC nanoparticles, *Materials Science & Engineering A* 909 (2024) 146861. <https://doi.org/10.1016/j.msea.2024.146861>.
- [18] Y. Jiang, J. Xi, L. Du, W. Lei, J. Xing, C. Liu, Q. Li, H. Gu, Time-varying electrochemically heterogeneous entities during the micro-galvanic corrosion associated with nanoscale precipitates in an Al–Cu–Li alloy, *Corrosion Science* 206 (2022) 110542. <https://doi.org/10.1016/j.corsci.2022.110542>.

- [19] X. Yang, G. Tian, D. Liu, B. Ren, H. Cai, M. Liu, Y. Zhang, Experimental study and 3D cellular automaton simulation of intergranular corrosion on laser-welded joints of Al-Cu-Li alloys under acidic conditions, *Defence Technology* 58 (2026) 314-329. <https://doi.org/10.1016/j.dt.2025.10.025>.
- [20] Z. Fu, X. Li, Cirenpuchi, J. Jin, L. Liao, Corrosion and fatigue behavior of a stress-aged 2195 Al-Cu-Li alloy, *Engineering Failure Analysis* 172 (2025) 109418. <https://doi.org/10.1016/j.engfailanal.2025.109418>.
- [21] H. Song, J. Liu, H. Zhang, J. Du, Multi-source data driven fatigue failure analysis and life prediction of pre-corroded aluminum–lithium alloy 2050-T8, *Engineering Fracture Mechanics* 292 (2023) 109626. <https://doi.org/10.1016/j.engfracmech.2023.109626>.
- [22] V.K. Yadav, V. Gaur, I.V. Singh, Effect of pre-corrosion on crack initiation behavior of friction stir welded aluminum alloy 2024-T3 in high and low cycle fatigue regimes, *International Journal of Fatigue* 179 (2024) 108037. <https://doi.org/10.1016/j.ijfatigue.2023.108037>.
- [23] W. Wang, H. Li, X. Ruan, Y. Wang, W. Zhao, Q. Wang, H. Chen, Study on corrosion fatigue performance of aluminum alloy high power laser filled wire welding joints, *Engineering Failure Analysis* 191 (2026) 110781. <https://doi.org/10.1016/j.engfailanal.2026.110781>.
- [24] C. Wang, Z. Jiang, Y. Zhang, J. Huang, F. Han, L. Xue, Y. Zhang, Precipitate-mediated mechanical-corrosion property trade-offs in TIG-welded Al-3.2Cu-1.48Li alloy joint, *Corrosion Science* 256 (2025) 113197. <https://doi.org/10.1016/j.corsci.2025.113197>.
- [25] H. Song, H. Shao, Y. Zhou, C. Zhao, J. Chi, G. Dong, J. Bi, Q. Zheng, X. Jia, M.D. Starostenkov, C. Tan, Effect of TiC ceramic particles and circular oscillating laser beam on corrosion resistance in fusing zone of 2195 Al-Li alloy, *Corrosion Science* 252 (2025) 112986. <https://doi.org/10.1016/j.corsci.2025.112986>.
- [26] X. Wang, G. Li, Q. He, J. Jiang, D. Li, W. Shao, L. Zhen, Intergranular corrosion of an Al-Cu-Li alloy: The influence from grain structure, *Corrosion Science* 211 (2023) 110845. <https://doi.org/10.1016/j.corsci.2022.110845>.
- [27] F. Wen, J. Chen, S. Han, J. Ye, J. Ren, Y. Zhang, W. Li, R. Guan, The initial corrosion behavior of Al-Cu-Li bicrystals: effect of misorientation and precipitates, *Journal of Materials Research and Technology* 18 (2022) 3716-3724. <https://doi.org/10.1016/j.jmrt.2022.04.024>.
- [28] Y. Li, Y. Wang, Y. Lv, Y. Li, B. Liu, Z. Dong, X. Zhang, Localized corrosion of friction stir welded Al–Cu–Li alloy: The influence of grain-stored energy on grain boundary micro-chemistry, *Journal of Materials Research and Technology* 37 (2025) 3880-3894. <https://doi.org/10.1016/j.jmrt.2025.07.073>.
- [29] C. Han, M. Yu, X. Jia, Z. Zhao, C. Zhang, J. Xiao, S. Li, J. Liu, Effect of inhomogeneous microstructures on stress corrosion cracking sensitivity of 2050-T84 Al-Li alloy thick plate, *Corrosion Science* 235 (2024) 112213. <https://doi.org/10.1016/j.corsci.2024.112213>.
- [30] Y. Zhang, Y. Lv, E. Liu, G. Cai, Q. Pan, B. Liu, Z. Dong, X. Zhang, Corrosion morphology evolution in 2A97-T6 Al-Cu-Li alloy: The influence of grain-stored energy, *Materials Characterization* 191 (2022) 112169. <https://doi.org/10.1016/j.matchar.2022.112169>.
- [31] Y. Lv, X. Meng, Z. Dong, Y. Ma, X. Zhang, Localized corrosion evolution associated with intermetallic particles in Al-Cu-Li alloys, *Materials Characterization* 204 (2023) 113144. <https://doi.org/10.1016/j.matchar.2023.113144>.
- [32] J. Zhao, Z. Dan, H. Chang, L. Zhou, Cracks propagation mechanism of intergranular corrosion in Al–Cu–Li AA2195 alloys governed by constant relative humidity, *Journal of Materials Research and Technology* 32 (2024) 165-176. <https://doi.org/10.1016/j.jmrt.2024.07.145>.
- [33] C. Deng, C. Wang, F. Wang, B. Song, H. Zhang, Effect of microstructure evolution on corrosion behavior of 2195 Al-Li alloy friction stir welding joint, *Materials Characterization* 184 (2022) 111652. <https://doi.org/10.1016/j.matchar.2021.111652>.
- [34] K.-J. Song, Y.-H. Wu, P.-C. Liang, X. Fu, Z.-G. Hu, G.-Q. Chen, W.-L. Zhou, Microstructure and crack propagation behavior of 2195 Al–Li alloy with different orientations in the friction stir welding nugget zone, *Materials Science & Engineering A* 920 (2025) 147537. <https://doi.org/10.1016/j.msea.2024.147537>.
- [35] L. Kou, Y. Zhao, G. Wang, X. Cheng, Y. Ouyang, Z. Yu, In situ study of micro-damage evolution in 2195

- aluminum alloy, *International Journal of Mechanical Sciences* 306 (2025) 110848. <https://doi.org/10.1016/j.ijmecsci.2025.110848>.
- [36] P. Wang, L. Ye, Z. Li, X. Guo, J. Tang, S. Liu, Y. Zhang, Effect of microstructure characteristics on fatigue properties of Al-Cu-Li alloy and damage accumulation mode of fatigue crack propagation under constant load, *Journal of Alloys and Compounds* 1038 (2025) 182978. <https://doi.org/10.1016/j.jallcom.2025.182978>.
- [37] H. Song, R. Liang, J. Ren, J. Liu, H. Zhang, Q. Zhang, Failure analysis and life prediction of pre-corroded 2050-T8 Al-Li alloy under in-phase multiaxial fatigue loading, *Engineering Failure Analysis* 180 (2025) 109881. <https://doi.org/10.1016/j.engfailanal.2025.109881>.
- [38] V.K. Yadav, V. Gaur, I.V. Singh, Corrosion-fatigue behavior of welded aluminum alloy 2024-T3, *International Journal of Fatigue* 173 (2023) 107675. <https://doi.org/10.1016/j.ijfatigue.2023.107675>.
- [39] Y. Chen, F. Kong, J. Yang, X. Song, X. Meng, Damage behavior of 2198-T8 Al-Li alloy with different corrosion fatigue modes, *International Journal of Fatigue* 156 (2022) 106671. <https://doi.org/10.1016/j.ijfatigue.2021.106671>.
- [40] H. Song, C. Liu, H. Zhang, X. Yang, Y. Chen, S.B. Leen, Experimental investigation on damage evolution in pre-corroded aluminum alloy 7075-T7651 under fatigue loading, *Materials Science & Engineering A* 799 (2021) 140206. <https://doi.org/10.1016/j.msea.2020.140206>.
- [41] Y. Chen, Y. Zhang, Z. Cui, F. Wang, T. Dong, X. Meng, Effect of impact-corrosion coupling damage on fatigue properties of 2198-T8 aluminum-lithium alloy, *Engineering Failure Analysis* 163 (2024) 108471. <https://doi.org/10.1016/j.engfailanal.2024.108471>.
- [42] N. Ao, S. Tang, X. Zhang, Y. Fu, S. Wu, 4D evolution of corrosion fatigue crack initiation and growth behavior in high-strength Al alloy by in situ synchrotron X-ray tomography, *Corrosion Science* 262 (2026) 113634. <https://doi.org/10.1016/j.corsci.2026.113634>.
- [43] N. Ao, S. Tang, Z. Zhang, H. Huang, Z. Chen, Y. Fu, K. Li, G. Kang, S. Wu, Tracing corrosion fatigue crack initiation in high-strength aluminum alloy, *International Journal of Mechanical Sciences* 318 (2026) 111579. <https://doi.org/10.1016/j.ijmecsci.2026.111579>.