



Article

Study on the Synergistic Effects of Pre-Deformation and Post-Aging Treatments on the Mechanical and Corrosion Properties of a 2A97 Al-Cu-Li Alloy

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Abstract

This work systematically investigates the effects of pre-deformation, post-aging temperature, and aging time on the mechanical properties, corrosion behavior, and microstructure of a 2A97 Al-Cu-Li alloy. Microstructural characterization indicates that the main precipitates are T_1 (Al_2CuLi), δ' (Al_3Li), θ' (Al_2Cu), and S' (Al_2CuMg). At 160 °C with 0% pre-deformation strain, increasing aging time increases the size and number density of T_1 , changing the size and number density of δ' and θ' accordingly, whereas raising the aging temperature to 180 °C mainly coarsens precipitates. Increasing pre-deformation from 0% to 12% increases T_1 number density and refines its size. Significantly, after aging at 180 °C, θ' is absent in the 8% and 12% pre-deformed alloys, and T_1 re-dissolves in the over-aged 12% alloy. Grain-boundary (GB) phases evolve from fine/discontinuous (under-aged) to fine/continuous (peak-aged) to coarse/discontinuous (over-aged). At the same time, higher pre-strain reduces their size, and higher aging temperature promotes coarsening. Higher aging temperature and pre-deformation accelerate age hardening and shorten the peak-aging time. In the peak-aged state, strength increases but elongation decreases with increasing pre-deformation or aging temperature. Among all, the 4% pre-deformed alloy aged at 160 °C for 30 h shows optimal room-temperature properties, i.e., ultimate tensile strength ~613 MPa, yield strength ~564 MPa, and total elongation ~6.91%. Corrosion resistance was closely related to microstructural evolution during aging, and both prolonged aging and increased pre-deformation decreased the susceptibility to intergranular corrosion. The results provide a basis for optimizing the thermomechanical processing of 2A97 Al-Cu-Li alloys.



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1. Introduction

Al-Li alloys have been widely used in aerospace engineering due to their high strength and low density [1,2]. Over the past decades, third-generation Al-Li alloys, in which Li is partially replaced by Cu, combined with microalloying with specific elements such as Mg and Ag, have rapidly advanced and exhibited superior performance, e.g., high deformability, low anisotropy, and exceptional fatigue resistance [3–5]. Among current third-generation Al-Li alloys, 2A97 Al-Cu-Li is a high-strength, age-hardenable alloy. It is

well recognized that the strength of the 2A97 alloy arises from the synergistic contribution of solid-solution, grain-boundary, work hardening, and, most importantly, precipitation strengthening [6,7]. In particular, precipitation strengthening commonly plays a critical role, and the precipitation scenario evolves with aging time [8]. For the 2A97 Al-Cu-Li alloy, fine coherent δ' phases are widely generated in the early aging stage [9]. They can be readily sheared by dislocations, increasing interfacial energy and impeding slip. As aging proceeds, the more stable T_1 and θ' phases become dominant, and the mechanism shifts to Orowan bypassing, in which dislocations form loops around non-shearable particles, significantly increasing strength [10–12]. The dense dispersion of T_1 plates is particularly effective. Moreover, aging temperature further controls precipitation behavior: lower temperatures slow nucleation and growth due to limited diffusion, while higher temperatures accelerate these processes but can lead to excessive coarsening and loss of ductility in the Al-Cu-Li alloy [13,14].

Regarding corrosion, the electrochemical behavior of each precipitate plays a decisive role. The T_1 phase initially acts as an anode (more negative than the Al matrix) and dissolves, but copper enrichment gradually shifts its potential to a cathodic value, causing the surrounding matrix to dissolve instead; this cyclic process sustains corrosion [15]. The θ' phase is cathodic relative to the matrix, promoting anodic dissolution of adjacent Al [16]. The δ' phase is anodic and tends to cause uniform corrosion, while the S' phase initially dissolves preferentially via magnesium depletion, after which the neighboring matrix becomes anodic [17,18]. Grain-boundary phases are electrochemically active and undergo a similar dynamic transformation [19]. Any change in the relative proportions of these precipitates alters local potential differences and, thus, the overall corrosion behavior. Therefore, aging conditions that can efficiently tune precipitate characteristics are essential to the overall performance of the 2A97 Al-Cu-Li alloy.

Dislocations are important crystalline defects that strongly influence precipitation processes in age-hardenable aluminum alloys. Owing to their associated strain fields and excess energy, dislocations can serve as preferential nucleation sites for precipitates and provide accelerated diffusion pathways for solute atoms [20]. Consequently, the nucleation, growth, and spatial distribution of precipitates are closely related to the dislocation density within the alloy. In Al-Cu-Li alloys, numerous studies have demonstrated that the T_1 phase preferentially nucleates on dislocations and subgrain boundaries, making precipitation kinetics particularly sensitive to thermomechanical processing [21–23]. Therefore, controlling the density and distribution of dislocations is an effective approach for tailoring precipitate evolution and the resulting mechanical properties of Al-Cu-Li alloys.

In addition, pre-deformation, applied after solution treatment, has also been reported to alter the precipitate characteristics of Al-Cu-Li alloys [24,25]. Generally, pre-deformation introduces a high density of dislocations and stored strain energy, which provide preferential nucleation sites for precipitates and promote the precipitation of the T_1 phase in Al-Cu-Li alloys, thereby enhancing age-hardening kinetics [10,26]. As the pre-deformation level increases, the T_1 plates become more uniformly distributed, increase in number density, and decrease in size, leading to a higher yield strength and a shorter time to reach peak aging [27]. However, as reported by Rodgers et al. [28], the beneficial effect has a threshold. Excessive deformation in Al-Cu-Li alloys leads to dislocation tangling and cell formation, hindering T_1 precipitation and potentially promoting microcrack initiation, thereby reducing damage tolerance. Additionally, sufficient pre-deformation suppresses θ' precipitation and narrows the precipitate-free zones (PFZs) along grain boundaries [29,30]. Therefore, pre-deformation should be carefully applied to achieve the strength and ductility synergy of Al-Cu-Li alloys.

Motivated by the above considerations, the practical industrial application of 2A97 alloy requires a thorough understanding of how aging heat treatment and pre-deformation synergistically affect its microstructure and mechanical properties. Aging parameters, such as temperature and time, along with pre-deformation amounts, directly influence precipitate characteristics, which, in turn, determine the alloy's performance [31,32]. As discussed previously, it is well established that the strength of Al-Cu-Li alloys primarily depends on intragranular T_1 and θ' phases. At the same time, grain-boundary precipitates and PFZs largely govern corrosion resistance. Since these microstructural features can be tailored by adjusting process conditions, studying the relationships among aging, pre-deformation conditions, and the resulting microstructure and properties is essential for developing a 2A97 alloy that offers both high strength and good corrosion resistance. However, the above issues lack in-depth investigation. Moreover, how pre-deformation and post-aging conditions determine the microstructural features (especially the precipitate profiles), thereby governing mechanical performance and corrosion resistance, remains unclear.

Unlike previous studies [20,33] that mainly focused on either aging behavior or mechanical properties, the present work combines microstructural characterization, mechanical testing, intergranular corrosion evaluation, and electrochemical measurements to establish a comprehensive relationship between precipitate evolution and the resulting property changes in 2A97 alloy. Particular attention is given to how pre-deformation and subsequent aging modify both intragranular and grain-boundary precipitates, thereby governing the strength and corrosion resistance of the alloy.

Therefore, this work investigates the influence of pre-deformation and post-aging conditions on the microstructure, strength, and corrosion resistance of the 2A97 alloy. Using a combination of testing and characterization techniques, including hardness testing, room-temperature tensile testing, intergranular corrosion testing, and electrochemical measurements, the correlation between microstructure and properties is explored. The aim is to identify aging regimes that achieve a favorable balance between high strength and corrosion resistance, and to systematically elucidate the underlying mechanisms by which the microstructure governs these properties.

2. Materials and Experimental Details

2.1. Material and Specimen Preparation

Raw materials for 2A97 Al-Cu-Li alloy plates were purchased from Southwest Aluminum (Group) Co., Ltd., (Chongqing, China) with the actual chemical composition measured by an Agilent 5110 Vertical Dual View ICP-OES (Agilent Technologies, Santa Clara, CA, USA), as listed in Table 1. A large plate (600 mm × 200 mm × 2 mm) was first subjected to solid-solution heat treatment at 520 °C for 90 min in a furnace, then water-quenched to room temperature. After that, the plate was cut into several small pieces (200 mm × 100 mm × 2 mm), and these pieces were then cold-rolled by different strains from 4% to 12% (see Figure 1a). To facilitate the description of different thermomechanical conditions, a sample designation system was adopted, as summarized in Table 2. In this notation, P4-160-30 denotes a specimen subjected to 4% pre-deformation and subsequently aged at 160 °C for 30 h. The designation T6 refers to the temper condition obtained by solution heat treatment followed by artificial aging. Pre-deformation was introduced by cold rolling using a CNZW-301C two-high rolling mill with roll dimensions of $\Phi 300 \times 400$ mm (Changchun Research Institute for Mechanical Science Co., Ltd., Changchun, China). The solution-treated specimens with an initial thickness of 2.0 mm were rolled in a single pass to thicknesses of 1.92, 1.84, and 1.76 mm, corresponding to thickness reductions of 4%, 8%, and 12%, respectively. The rolling was performed at room temperature prior to the subsequent aging treatment. Aging heat treatments were performed on the pre-strained

specimens at 160 °C or 180 °C for a series of durations from 2 to 300 h, followed by water quenching. The selected aging times were intended to cover the under-aged, peak-aged, and over-aged conditions of the alloy. The aging heat treatments were carried out in an electric heating blast-drying oven. Before each aging treatment, the drying oven temperature was calibrated, and the temperature variation was kept within ± 1 °C. Specimens for hardness and corrosion tests were cut using wire electrical discharge machining. Hardness specimens with dimensions of $10 \times 10 \times 2$ mm were prepared parallel to the rolling surface.

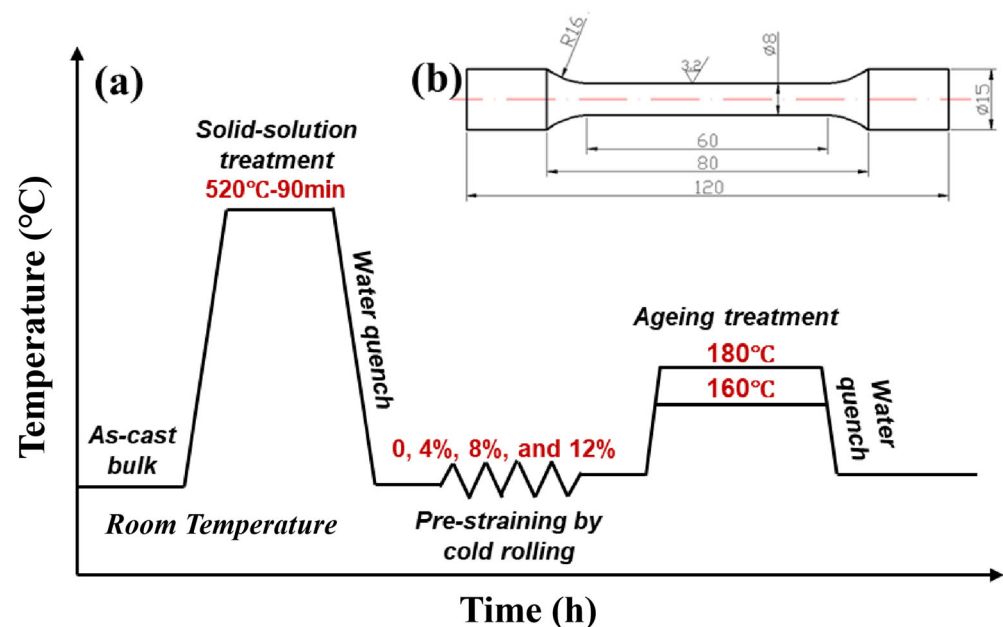


Figure 1. (a) The illustration of the processing routines of the 2A97 Al-Cu-Li alloys in this work. (b) The dimension of the wire-cut bar-shaped specimens used in the room-temperature tensile experiments.

Table 1. The actual chemical composition of the 2A97 Al-Cu-Li alloy (in wt.%) measured by ICP-OES.

Elements	Cu	Li	Mg	Mn	Zn	Zr	Ti	Fe	Si	Al
Wt.%	3.74	1.48	0.46	0.29	0.47	0.12	0.04	0.05	0.02	Bal.

Table 2. Summary of sample designations and processing conditions used in this study.

Sample Designation	Pre-Deformation (%)	Aging Temperature (°C)	Aging Time (h)
ST	0	—	—
P0-160-t	0	160	t
P4-160-t	4	160	t
P8-160-t	8	160	t
P12-160-t	12	160	t
P0-180-t	0	180	t
P4-180-t	4	180	t
P8-180-t	8	180	t
P12-180-t	12	180	t

2.2. Mechanical and Hardness Characterization

The hardness test was conducted on an HV-50 Vickers hardness tester (Laizhou Huayin Testing Instrument Co., Ltd., Laizhou, China), applying a 500 g load and holding it for 15 s. Before each hardness test, the specimen surface was sequentially polished with 80-, 400-, 800-, 1200-, and 2000-grit sandpaper, then polished with 0.5-micron diamond paste, and finally cleaned with ethanol and dried with cold air. To ensure the accuracy of the hardness

data of the specimen, more than 7 hardness values were measured simultaneously within the uniformly spaced test points, and the average value and error bars were calculated.

In addition to the hardness tests, uniaxial tensile tests were conducted on these pre-strained and aged alloys to evaluate mechanical properties better and determine the peak-ageing parameters. The tensile bar-shaped specimens were wire-cut at the half-thickness position of the plate along the initial rolling direction into the cylindrical tensile specimens shown in Figure 1b. The length of the parallel section of the bar-like specimen was ~60 mm, and the diameter of the gauge section was ~8 mm. The tensile specimen geometry was designed according to the dimensions of the processed sheet and the requirements of this study. The tensile properties were evaluated using an MTS-810 servo-hydraulic testing machine (MTS Systems Corporation, Eden Prairie, MN, USA) at room temperature. The tests were performed under displacement-controlled mode at a constant crosshead speed of 2 mm/min, corresponding to quasi-static loading conditions. The strain was measured precisely with a 25 mm extensometer. Three parallel specimens were prepared for each batch of aged alloys, and the tensile, yield, and elongation strengths obtained from the parallel experiments were averaged to improve reliability.

2.3. Corrosion Characterization

Intergranular corrosion (IGC) tests and electrochemical measurements were conducted to evaluate the corrosion behavior of the 2A97 Al-Cu-Li alloy. The intergranular corrosion (IGC) test was conducted on the initial rolling surface of sheet-like specimens measuring approximately 20 mm × 20 mm × 2 mm. The other five non-test surfaces of each sheet-like specimen were sealed with epoxy resin. Before the IGC test, the test surface was ground with sandpaper, then polished with diamond gypsum, and subsequently cleaned with ethanol. After drying, the sheets were first immersed in a 10% NaOH solution for 5 min, then immediately immersed in a 30% HNO₃ solution until the black products on the test surface disappeared. The residual solution was washed with clean water, then immersed in a mixed solution of 57 g/L NaCl and 10 mg/L H₂O₂ for 6 h. The IGC test was carried out in an electrically heated constant-temperature water bath at 35 ± 2 °C. After the long-term immersion, the surface corrosion products were washed with a dilute nitric acid solution. One third of the specimen was cut along a direction perpendicular to the rolling direction, and the cross-section was carefully ground. The corrosion damage was characterized using a Leica metallographic microscope at a magnification of 200×. For each specimen, at least 20 images of typical corrosion cross-sections were obtained from each test specimen using an optical microscope to facilitate subsequent statistical analysis of corrosion morphologies.

Electrochemical tests were also conducted on these aged specimens using a CHI-660B electrochemical workstation (Shanghai Chenhua Instrument Co., Ltd., Shanghai, China). The polarization curves and open-circuit potentials (OCPs) of the specimens were obtained. The electrochemical test was carried out using the three-electrode test system of the electrochemical workstation, where the working electrode was the tested specimen, the saturated calomel electrode (SCE) was used as the reference electrode, and the platinum sheet material with a size of around 15 mm × 15 mm was used as the auxiliary electrode. The specimen was cut into a 20 mm × 20 mm × 2 mm sheet-like piece, the top surface was polished, and the sheet was then subjected to a room-temperature electrochemical test. The test solution was 3.5 wt% NaCl. The exposed area of the tested specimen was ~1 cm². During the electrochemical test, the aluminum-copper-lithium alloy surface typically exhibited high electrochemical activity. Therefore, the specimen was connected to the three-electrode test system to stabilize for 10 min before measuring its open-circuit potential. The initial test time was set at 600 s. When the potential fluctuated less than 2 mV within 1 min, it was recorded as the open-circuit potential. Otherwise, the test time

was further extended. Then, the open-circuit potential range of +600 mV to +800 mV was selected to test the polarization curve, and the scan rate was set to 1 mV/s. The polarization curve was analyzed and fitted using the CHI660E software electrochemical workstation software (CH Instruments, Austin, TX, USA).

2.4. Microstructural Characterization

The microstructural features of the as-aged and post-tested specimens were characterized using a Sirion 200 field-emission scanning electron microscope (FEI Company, Hillsboro, OR, USA). The distribution of second-phase particles in the alloy was observed in the backscattered electron mode. The fracture morphology of the specimens after tensile testing was observed using secondary electron imaging of the SEM. The instrument accelerating voltage was 25 kV, and the scanning step size was 1 μm . Microstructural characterization at sub-micro scales was carried out using a Tecnai G2 F20 transmission electron microscope (FEI Company, Hillsboro, OR, USA) at an accelerating voltage of 200 kV. For TEM examination, the specimens were first cut into 5 mm $\times$ 2 mm $\times$ 2 mm and thinned to 80–100 μm using sandpaper, then punched into a circular piece with a diameter of 3 mm. The circular thin slice was thinned using an MPT-1 double-spray electrolytic thinning instrument (Shanghai Jiaoda Electromechanical Technology Development Co., Ltd., Shanghai, China). The used double-spray liquid was a mixture of nitric acid and methanol at a volume ratio of 3:7. In the double-spray chamber, liquid nitrogen was used for cooling, reducing the working temperature to $-25\text{ }^{\circ}\text{C}$. The constant current method was used, with the current set to 80–90 mA and the voltage to 15–25 V. TEM foils for microstructural characterization were prepared from the longitudinal section parallel to the rolling direction (RD).

3. Experimental and Characterization Results

3.1. Mechanical Properties of the Pre-Strained and Aged Specimens

First, the hardness versus aging time curves for the 2A97 Al-Li alloy aged at $160\text{ }^{\circ}\text{C}$ are shown in Figure 2a. It is found that alloys subjected to different pre-straining exhibit similar age-hardening characteristics. As aging time increases, the hardness of alloys with different pre-strains first increases rapidly, reaches a peak, remains stable for a period, and then declines slowly. P0-160-t alloy exhibits the lowest hardness. After aging for 12 h, this alloy remains under-aged, with a hardness of $\sim 141\text{ HV}$. Its hardness reaches the peak-aged value of $\sim 183\text{ HV}$ after 78 h. For the P4-160-t specimen, the hardness increases significantly, and the age-hardening rate is also enhanced. The alloy reaches the peak-aged state after only 30 h, with a hardness value of 191 HV. Compared with the P0-160-t alloy, the time to reach peak aging is reduced by $\sim 48\text{ h}$, while the hardness increases by $\sim 8\text{ HV}$. Significantly, for the 8% and 12% pre-strain conditions, the aging time to peak hardness is reduced to 15 h and 12 h, respectively, with peak hardness values of 193 HV and 194 HV, respectively. The above results suggest that the larger the pre-strain applied before aging, the shorter the time required to reach the peak-aged state. Moreover, a distinct hardness plateau, in which the hardness variation does not exceed 5 HV around the peak hardness, is observed in these specimens. Additionally, the duration of such a peak hardness plateau decreases with pre-strain. The alloy free of pre-deformation exhibits the longest plateau period, ranging from 78 h to 200 h, while the alloy with 12% pre-strain shows the shortest plateau period, from 12 h to 48 h.

The room-temperature tensile properties versus aging time curves of the 2A97 Al-Cu-Li alloy with different pre-strains are presented in Figure 2b–d. It is evident that at the early stage of aging at $160\text{ }^{\circ}\text{C}$, the ultimate tensile strength (UTS) and yield strength (YS) of the alloy increase rapidly with aging time, reach a stable plateau, and then decrease to varying

degrees with further aging. Similar to the hardness test results, the aging-hardening rates obtained from the tensile experiments also differ among alloys with varying pre-strain values. As the pre-strain increases from 0% to 12%, the aging-hardening rate accelerates, the time to reach the peak strength shortens, and the duration of the peak plateau decreases. Figure 2d shows that the alloy's elongation decreases rapidly at first, then declines more slowly after reaching the peak aging time. Notably, when the applied aging time is held constant, the alloy without pre-deformation exhibits the highest elongation. As shown in the above results, after aging at 160 °C, pre-straining significantly accelerates the aging hardening of the 2A97 Al-Cu-Li alloy, leading to a faster attainment of the peak-aged state with increased peak hardness and strength, but also shortens the peak plateau duration and reduces elongation. As the pre-strain increases from 0% to 12%, the strength gradually rises while the elongation decreases. However, when the pre-strain exceeds 8%, further improvement in peak strength becomes very limited. Considering both strength and ductility, a pre-deformation of 4% followed by aging at 160 °C for 30–85 h provides a favorable combination of mechanical properties, with an average UTS of 613 MPa, a YS of 564 MPa, and a total elongation of ~6.9%.

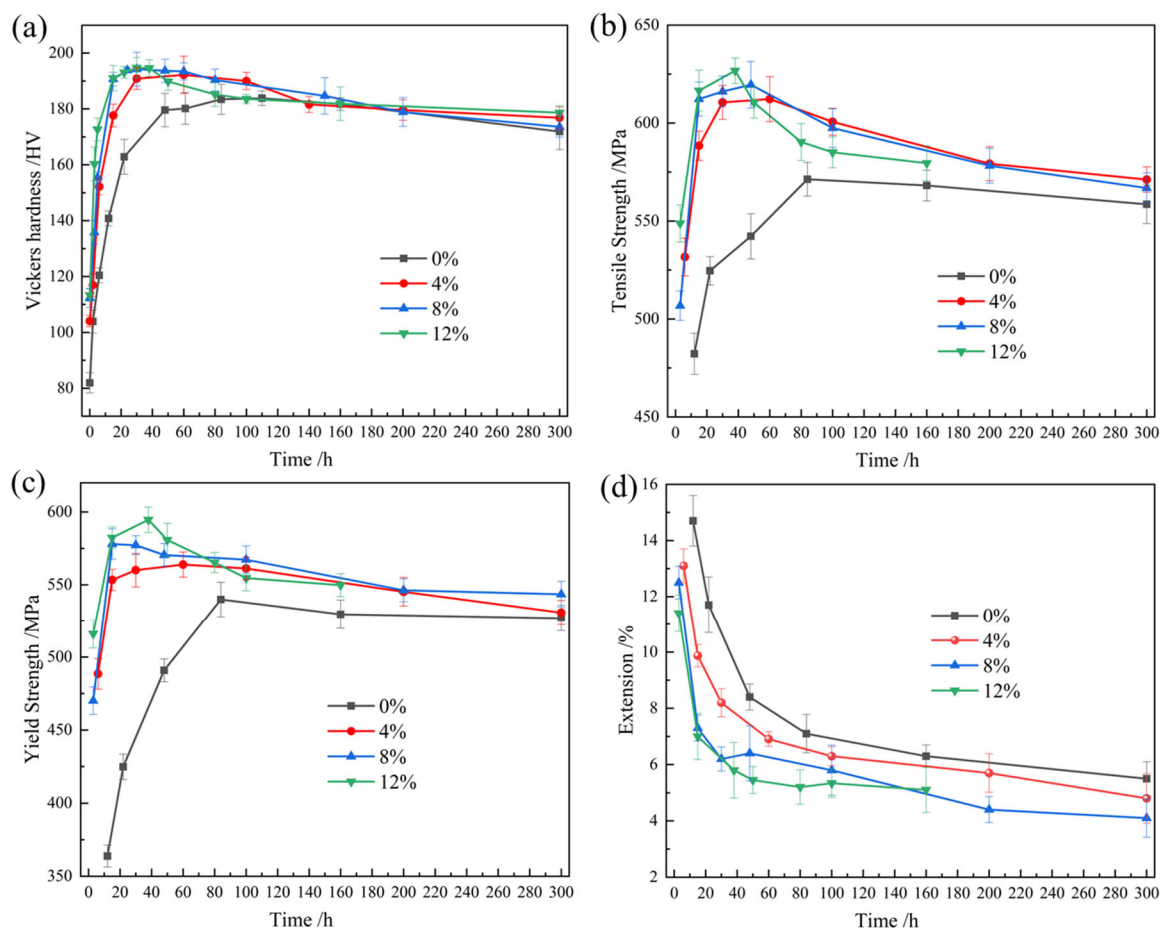


Figure 2. Hardness and tensile properties of 2A97 Al-Li alloy with different pre-strains aged at 160 °C: (a) hardness; (b) ultimate tensile strength; (c) yield strength; (d) elongation.

Figure 3a presents the age-hardening curves of the 180 °C-aged 2A97 Al-Li alloy with varying pre-strains. Similar to the 160 °C aging behavior, the hardness first increases rapidly to a peak, then remains stable for a period before slowly declining. However, the aging response is notably accelerated at 180 °C. The peak hardness of the 0% pre-strained alloy remains approximately 183 HV, while 4% pre-deformation raises it to about 192 HV.

Further increasing the pre-deformation yields only marginal increases in hardness (<4 HV), consistent with the trend observed at 160 °C. Compared with aging at 160 °C, the time to reach the peak-aged state at 180 °C is significantly reduced. For the 4% pre-deformed alloy, the peak-aged stage shifts from 30–85 h at 160 °C to 10–26 h at 180 °C, a reduction of 20 h. For the 0% pre-deformed alloy, the reduction is even more pronounced (~54 h). In general, the smaller the pre-deformation, the greater the reduction in peak aging time when the temperature is raised to 180 °C. The room-temperature tensile properties at 180 °C are shown in Figure 3b–d. The strength increases rapidly to a peak, then plateaus, and then gradually declines. Higher pre-deformation accelerates the aging hardening, shortens the time to peak strength, and reduces the duration of the peak plateau. As shown in Figure 3d, the elongation drops sharply in the early aging stage, then decreases more slowly during the peak plateau, and finally stabilizes. For the 0% pre-strained alloy, the peak-aged stage occurs between 24 h and 50 h, with an ultimate tensile strength of ~578 MPa, a yield strength of ~544 MPa, and a total elongation of ~7.0%. For the 4% pre-strained specimens (aging time from 10 h to 26 h), the corresponding values are ~625 MPa, ~580 MPa, and ~6.68%. For 8% (aging time from 5 h to 20 h), the values are ~628 MPa, ~584 MPa, and ~6.2%. And for the 12% (4–17 h): ~635 MPa, ~595 MPa, and ~5.9%. Overall, the pre-strained alloy aged at 180 °C exhibits favorable mechanical properties compared with all 160 °C-aged counterparts.

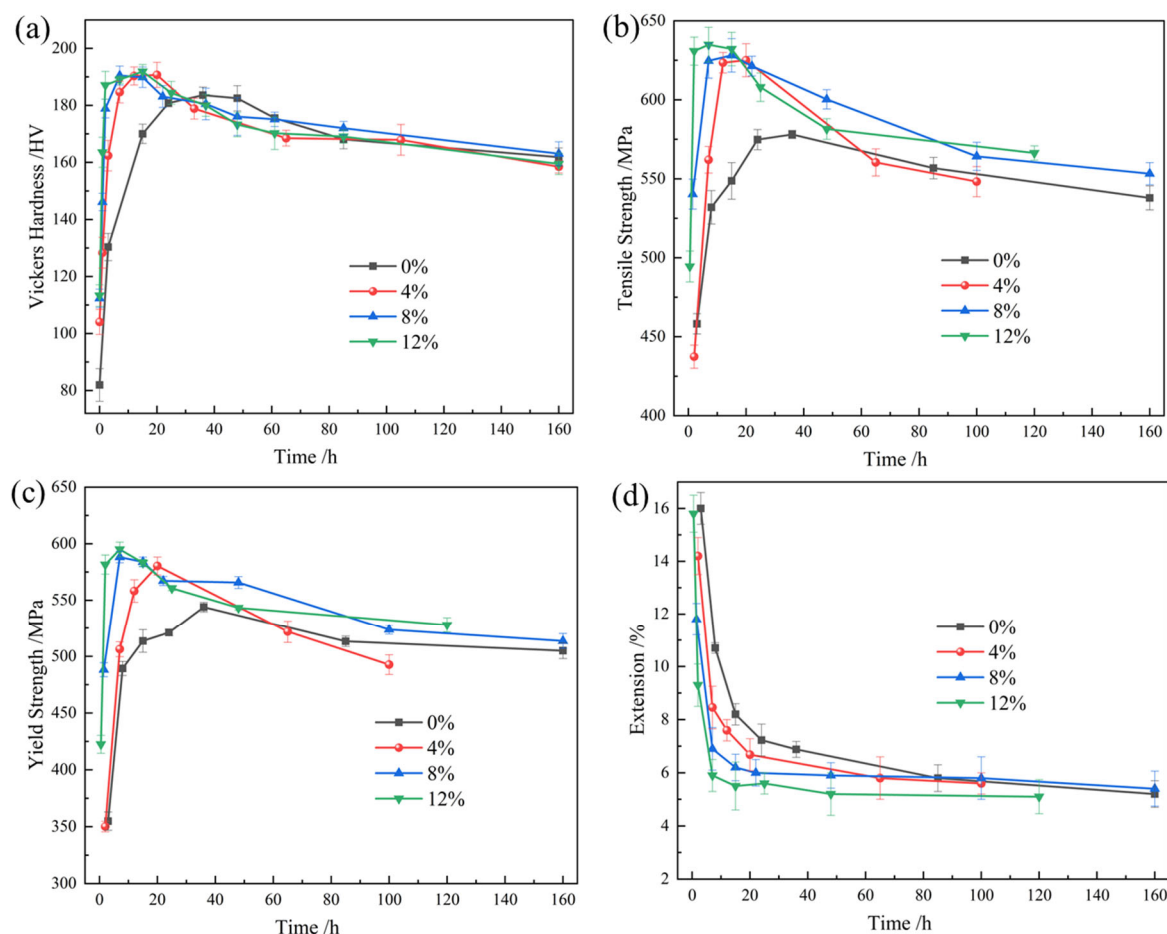


Figure 3. Hardness and tensile properties of 2A97 Al-Li alloy with different pre-strains aged at 180 °C: (a) hardness; (b) ultimate tensile strength; (c) yield strength; (d) elongation.

3.2. Corrosion Behaviors of the Pre-Strained and Aged Specimens

We first investigated the intergranular corrosion response of the pre-strained and aged specimens via immersion tests. It is well recognized that the intergranular corruptions observed in Al-Li alloys are commonly classified into four main types: (i) General intergranular corrosion (GIGC), where grid-like intergranular corrosion features cover more than 80% of the corroded area; (ii) Localized intergranular corrosion (LIGC), where such grid-like features occupy less than 80% of the corroded area; (iii) Pitting, characterized by small corrosion pits on the metal surface with cross-sectional views revealing cavities of varying sizes; and (iv) Pitting with local intergranular corrosion (P&LIGC), where slight intergranular corrosion appears around the bottom of pits, typically presenting shallow cracks and relatively large pits. The typical corrosion morphologies of the 2A97 Al-Li alloy with different pre-strains aged at 160 °C for various times are shown in Figure 4, and the corresponding statistical data on corrosion depth are listed in Table 3. The maximum corrosion depth of each corrosion type was measured to evaluate the alloy's resistance. As illustrated, the dominant corrosion modes for the differently aged alloys at 160 °C are LIGC and pitting. From the perspective of intergranular corrosion, LIGC occurs only in the 0% pre-strained alloy (i.e., the T6 state) in the under-aged (UA) condition, with a maximum depth of ~66.15 µm. With the pre-straining, the susceptibility of the aged alloys to intergranular corrosion decreases, and the corrosion mode shifts from LIGC to pitting. As the pre-strain level increases from 4% to 12%, the maximum pitting depth gradually rises from ~58.59 µm to ~77.73 µm. For the alloy without pre-strains, the dominant corrosion mode changes to pitting in the peak-aged (PA) and over-aged (OA) stages, with maximum pitting depths of ~83.48 µm and ~93.62 µm, respectively. This indicates that as aging time increases, resistance to intergranular corrosion increases. Moreover, from the perspective of pitting resistance, at the peak-aged state, the maximum pitting depths for alloys with 0%, 4%, 8%, and 12% pre-strains are ~83.48 µm, ~61.21 µm, ~75.73 µm, and ~86.21 µm, respectively, showing an initial decrease followed by an increase. P4-160-t alloy exhibits the smallest pitting depth. With extended aging time, the maximum pitting depth increases gradually for all specimens. Consequently, for the alloy aged at 160 °C, prolonged aging reduces pitting resistance, whereas increasing pre-strain from 0% to 12% first improves and then worsens pitting resistance.

Table 3. The summary of the corrosion mode and the maximum pitting depth of the 160 °C-aged specimens with different pre-strains.

Aging Conditions	DCM	MID (µm)	Aging Conditions	DCM	MID (µm)
P0-160-22 (UA)	LIGC	66.15	P8-160-3 (UA)	Pitting	60.23
P0-160-84 (PA)	Pitting	83.48	P8-160-48 (PA)	Pitting	75.73
P0-160-300 (OA)	Pitting	93.62	P8-160-150 (OA)	Pitting	92.81
P4-160-6 (UA)	Pitting	58.59	P12-160-3 (UA)	Pitting	77.73
P4-160-90 (PA)	Pitting	61.21	P12-160-38 (PA)	Pitting	86.21
P4-160-190 (OA)	Pitting	82.36	P12-160-100 (OA)	Pitting	120.42

Note: DCM denotes the dominant corrosion mode, while MID stands for the maximum intergranular corrosion depth here.

Furthermore, the typical corrosion morphologies of the alloy with different pre-deformation amounts aged at 180 °C are presented in Figure 5, and the corresponding corrosion depth statistics are given in Table 4. The dominant corrosion modes in the 180 °C-aged specimens remain LIGC and pitting. Regarding intergranular corrosion in the UA condition, LIGC is observed for alloys with 0%, 4%, and 12% pre-deformation, yielding the maximum depths of ~72.42 µm, ~48.34 µm, and ~78.16 µm, respectively. In contrast, it is noted that the 8% pre-strained alloy exhibits significant pitting, with a maximum depth of ~59.51 µm. As the pre-deformation increases from 0% to 12%, the corrosion mode transitions in the 180 °C-

aged specimens from LIGC to pitting and then back to LIGC, and the maximum intergranular corrosion depth first decreases and then increases, indicating that the susceptibility to intergranular corrosion first decreases and then rises. Moreover, raising the aging temperature from 160 °C to 180 °C increases the likelihood of LIGC, i.e., the alloy becomes more susceptible to intergranular corrosion. From the perspective of pitting resistance at the peak-aged state, the maximum pitting depths for alloys with 0%, 4%, 8%, and 12% pre-deformation again exhibit an initial decrease followed by an increase, with values of ~80.08 μm , ~62.41 μm , ~77.36 μm , and ~86.52 μm , respectively. The P4-180-t alloy has the lowest pitting depth. With increasing aging time, the maximum pitting depth increases for all specimens. As seen from the above results, it is inferred that an appropriate pre-deformation can efficiently slow down the pitting process. In summary, an appropriate pre-strain applied before aging reduces the susceptibility of the 2A97 Al-Li alloy to both intergranular corrosion and pitting. Comparing the data in Tables 3 and 4, it is found that increasing the aging temperature from 160 °C to 180 °C increases the alloy's susceptibility to intergranular corrosion. In contrast, prolonged aging time decreases intergranular corrosion susceptibility but accelerates pitting.

Table 4. The summary of the corrosion mode and the maximum pitting depth of the 180 °C-aged specimens with different pre-strains.

Aging Conditions	DCM	MID (μm)	Aging Conditions	DCM	MID (μm)
P0-180-3 (UA)	LIGC	72.42	P8-180-1.5 (UA)	Pitting	59.51
P0-180-36 (PA)	Pitting	80.08	P8-180-10 (PA)	Pitting	77.36
P0-180-85 (OA)	Pitting	84.13	P8-180-48 (OA)	Pitting	90.23
P4-180-2 (UA)	LIGC	48.34	P12-180-0.5 (UA)	LIGC	78.16
P4-180-12 (PA)	Pitting	62.41	P12-180-7 (PA)	Pitting	86.52
P4-180-65 (OA)	Pitting	81.88	P12-180-48 (OA)	P&LIGC	LIGC < 5 Pitting: 119.96

Electrochemical measurements were also conducted on the pre-strained and aged specimens, using a three-electrode test setup on the alloy immersed in a 3.5% NaCl solution. Figure 6 shows the Tafel polarization curves and open-circuit potential (OCP) curves of the 160 °C-aged alloy with different pre-strains (0%, 4%, 8%, and 12%) for various times. The corresponding electrochemical parameters are summarized in Table 4. Our analysis of the Tafel polarization curves shows that, for both anodic and cathodic polarization, the specimens under different aging conditions exhibit similar characteristics. During anodic polarization, all specimens exhibit an insignificant passivation stage, attributed to the formation of only a small protective oxide film on the alloy surface during anodic dissolution. From the Tafel curve fitting, the self-corrosion potential (E_{corr}) and self-corrosion current density (I_{corr}) can reflect the corrosion trend: a lower E_{corr} indicates higher corrosion susceptibility. In contrast, a higher I_{corr} suggests a faster corrosion propagation rate.

As shown in Table 5, for the P0-160-22 (UA) alloy, the I_{corr} is the lowest ($1.89 \mu\text{A}\cdot\text{cm}^{-2}$), and the E_{corr} is the most positive (-659.5 mV), indicating the best corrosion resistance. As the pre-deformation increases to 12%, I_{corr} gradually rises to $2.64 \mu\text{A}\cdot\text{cm}^{-2}$, and the corrosion resistance decreases. Moreover, with prolonged aging time, I_{corr} increases, and E_{corr} decreases, indicating an accelerated corrosion propagation rate and reduced corrosion resistance. As the aging time increases, the OCP first decreases rapidly and then remains relatively stable, as shown in Table 5. The OCP can preliminarily reflect the characteristics of intragranular precipitates, since the volume fraction of grain boundary phases is negligible compared to intragranular precipitates. The rapid initial drop in OCP is related to the precipitation of many phases (probably including the T_1 and δ' phases, widely observed in aged Al-Cu-Li alloys) that are more negative than the matrix. Additionally, it is observed

that the OCP and E_{corr} follow similar trends. At the same aging time, increasing pre-deformation leads to lower OCP and E_{corr} values, but higher I_{corr} . Therefore, the 0% pre-deformed alloy exhibits the best corrosion resistance. With extended aging time and increased pre-deformation, the alloy's corrosion resistance deteriorates, which is probably due to the formation of certain precipitates under higher chemical activity and greater sensitivity to corrosive liquids.

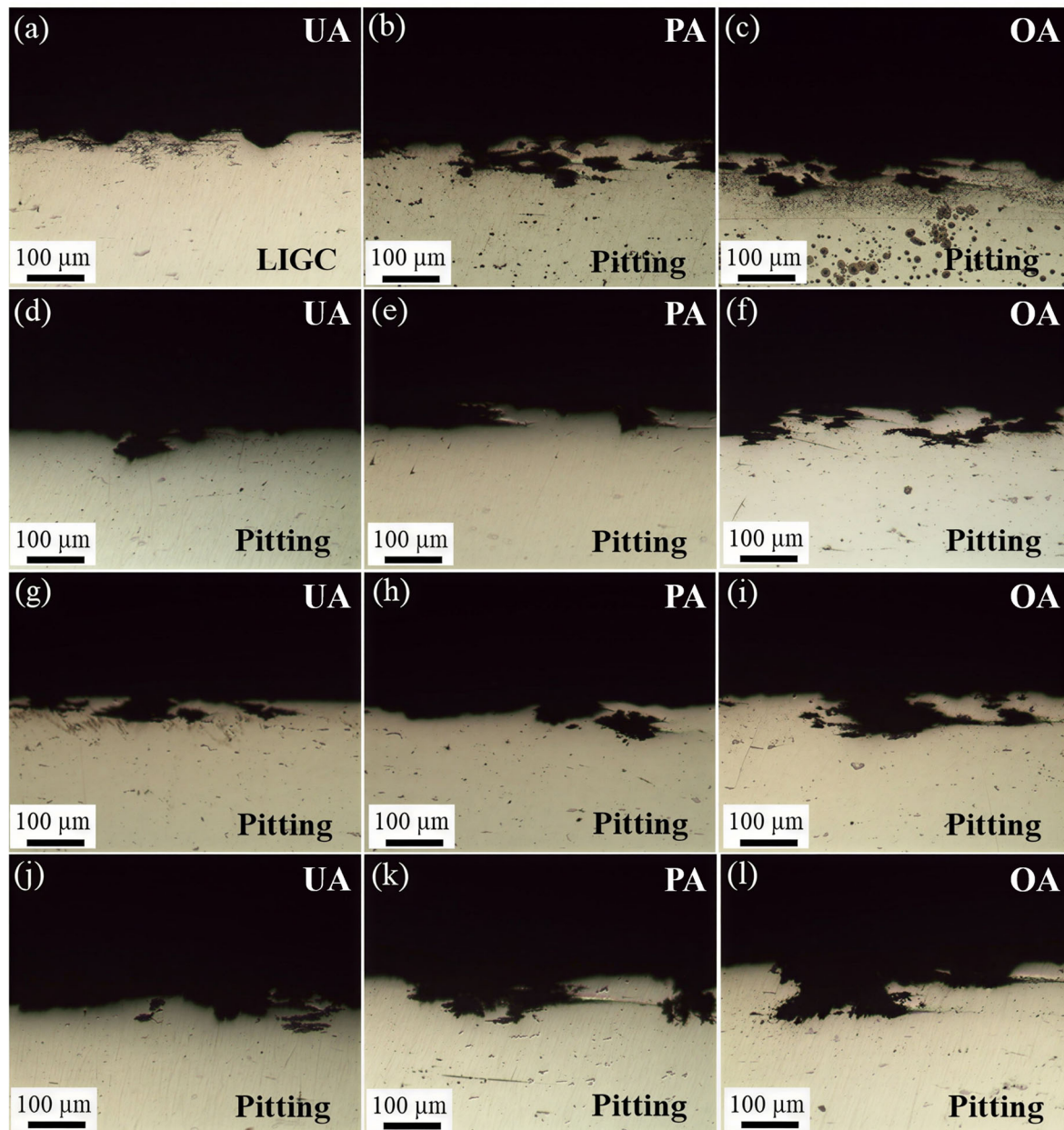


Figure 4. The typical corrosion morphologies of the 2A97 Al-Li alloy with different pre-strains aged at 160 °C for various times: (a) P0-160-22 (UA), LIGC; (b) P0-160-84 (PA), Pitting; (c) P0-160-300 (OA), Pitting; (d) P4-160-6 (UA), Pitting; (e) P4-160-90 (PA), Pitting; (f) P4-160-190 (OA), Pitting; (g) P8-160-3 (UA), (h) Pitting; P8-160-48 (PA), Pitting; (i) P8-160-150 (OA), Pitting; (j) P12-160-3 (UA), Pitting; (k) P12-160-38 (PA), Pitting; (l) P12-160-100 (OA), Pitting.

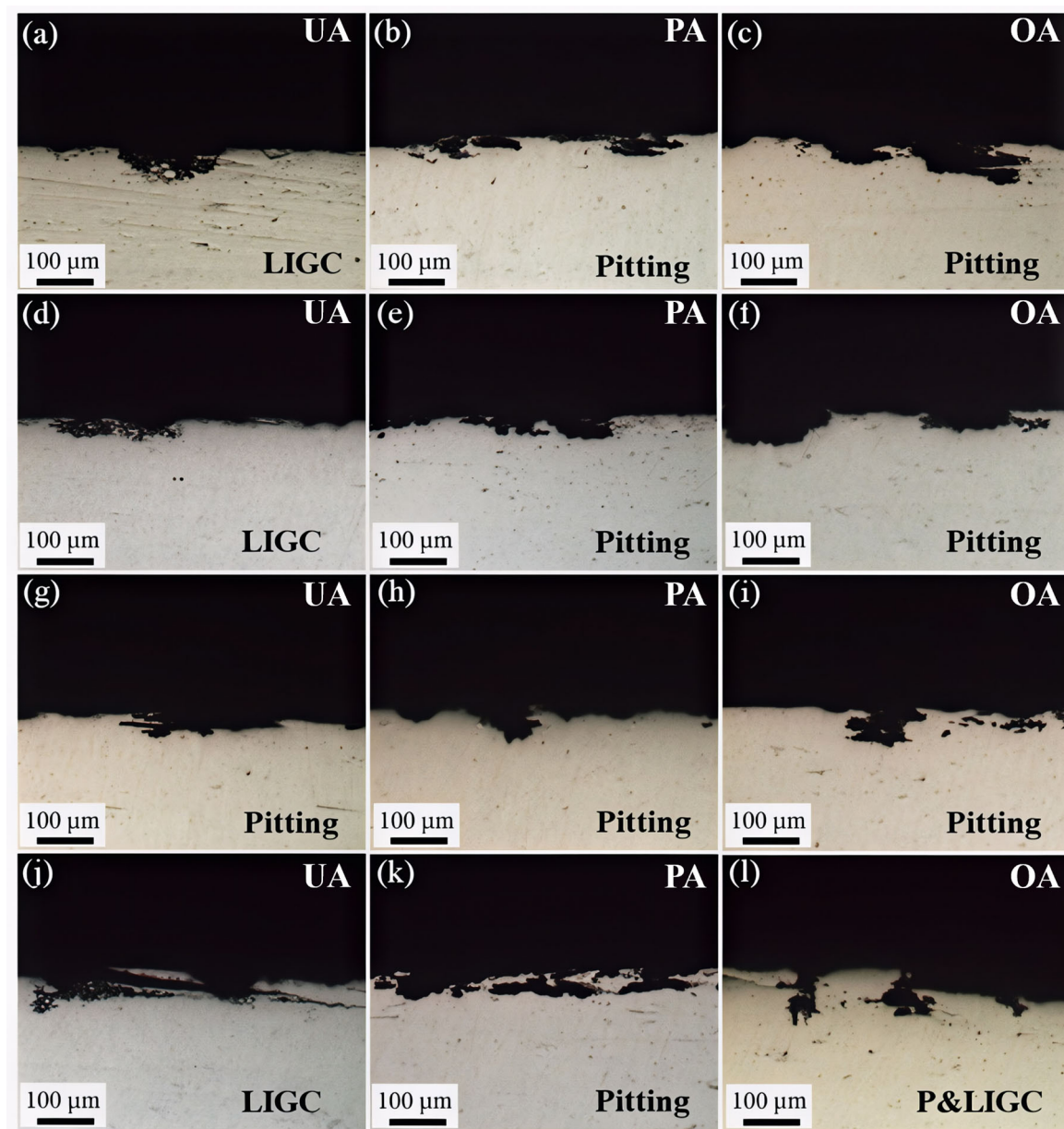


Figure 5. The typical corrosion morphologies of the 2A97 Al-Li alloy with different pre-strains aged at 180 °C for various times: (a) P0-180-3 (UA), LIGC; (b) P0-180-36 (PA), Pitting; (c) P0-180-85 (OA), Pitting; (d) P4-180-2 (UA), LIGC; (e) P4-180-12 (PA), Pitting; (f) P4-180-65 (OA), Pitting; (g) P8-180-1.5 (UA), Pitting; (h) P8-180-10 (PA), Pitting; (i) P8-180-48 (OA), Pitting; (j) P12-180-0.5 (UA), LIGC; (k) P12-180-7 (PA), Pitting; (l) P12-180-48 (OA), P&LIGC.

Table 5. Electrochemical parameters of 2A97 Al-Cu-Li alloy subjected to different pre-deformation levels and aged at 160 °C for various times.

Aging Conditions	E_{ocp} (mV)	E_{corr} (mV)	I_{corr} ($\mu A \cdot cm^{-2}$)	Aging Conditions	E_{ocp} (mV)	E_{corr} (mV)	I_{corr} ($\mu A \cdot cm^{-2}$)
P0-160-22	−671.7	−659.5	1.89	8%/3 h	−700.5	−670.2	2.58
P0-160-84	−704.1	−685.5	2.66	8%/48 h	−717.3	−692.6	3.44
P0-160-300	−710.9	−692.3	3.23	8%/150 h	−714.8	−700.3	4.01
P4-160-6	−690.9	−664.3	2.31	12%/3 h	−701.1	−676.8	2.64
P4-160-60	−709.6	−687.1	3.09	12%/38 h	−715.5	−694.8	3.65
P4-160-190	−715.2	−699.8	3.74	12%/100 h	−719.5	−706.5	4.39

Figure 7 shows the representative Tafel polarization and OCP curves of the alloys with different pre-strains (0%, 4%, 8%, and 12%) aged at 180 °C for various times. The corresponding electrochemical parameters are listed in Table 6 for reference. Figure 7a,b shows that the polarization curves of specimens under different aging conditions exhibit similar features. During anodic polarization, the same passivation phenomenon observed at 160 °C occurs, indicating that the increased temperature does not alter the characteristics of the anodic dissolution process. As shown in Table 6, for the P0-180-36 (PA) alloy, the I_{corr} is $3.05 \mu\text{A}\cdot\text{cm}^{-2}$, and E_{corr} is -689.5 mV . As pre-deformation increases to 12%, I_{corr} gradually rises to $4.36 \mu\text{A}\cdot\text{cm}^{-2}$, indicating reduced corrosion resistance. Similarly, with prolonged aging, I_{corr} increases, and E_{corr} decreases, reflecting a faster corrosion propagation rate and lower corrosion resistance.

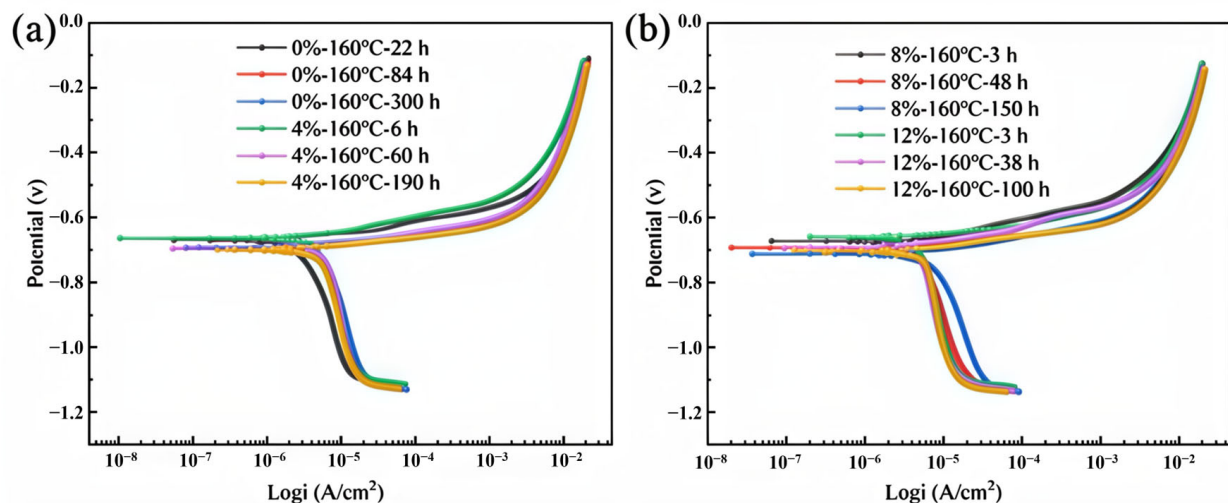


Figure 6. Typical polarization curves of pre-strained 2A97 Al-Cu-Li alloys at an aging temperature of 160 °C: (a) 0% and 4% pre-strain; (b) 8% and 12% pre-strain.

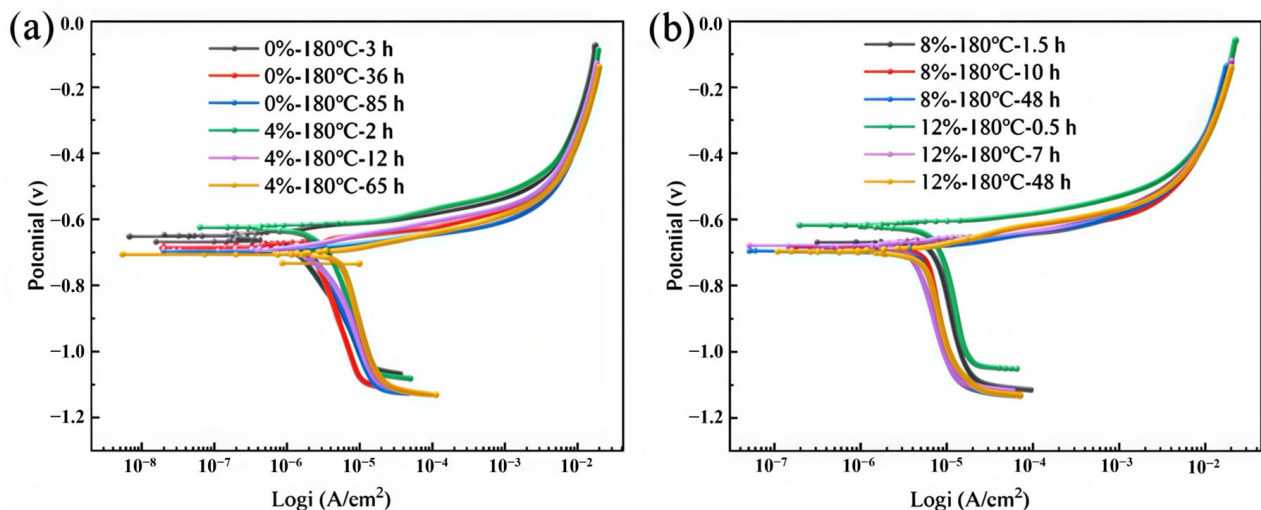


Figure 7. Typical polarization curves of pre-strained 2A97 Al-Cu-Li alloys at an aging temperature of 180 °C: (a) 0% and 4% pre-strain; (b) 8% and 12% pre-strain.

Again, as seen in Table 5, the OCP decreases rapidly at first and then remains relatively stable. Compared with the OCP curves at 160 °C, the OCP at 180 °C decreases more rapidly. Moreover, under the same pre-deformation and aging time, the alloy aged at 180 °C exhibits a more negative OCP, which is associated with the higher aging temperature, which promotes the nucleation and growth of T_1 phases. Moreover, among the 0% pre-

deformed alloys at different aging temperatures, it is observed that at the same aging time, higher aging temperatures result in lower E_{corr} and higher I_{corr} , indicating that corrosion resistance decreases with increasing temperature. In a word, the above electrochemical experimental results demonstrate that prolonging aging time, increasing aging temperature, and increasing pre-strain level all reduce the alloys' corrosion resistance.

Table 6. Electrochemical parameters of 2A97 Al-Cu-Li alloy subjected to different pre-deformation levels and aged at 180 °C for various times.

Aging Conditions	E_{ocp} (mV)	E_{corr} (mV)	I_{corr} ($\mu\text{A}\cdot\text{cm}^{-2}$)	Aging Conditions	E_{ocp} (mV)	E_{corr} (mV)	I_{corr} ($\mu\text{A}\cdot\text{cm}^{-2}$)
P0-180-3	−661.1	−651.1	1.61	8%/1.5 h	−698.2	−671.8	2.56
P0-180-36	−708.6	−689.5	3.05	8%/10 h	−717.5	−699.3	3.54
P0-180-85	−716.3	−700.5	4.14	8%/48 h	−721.5	−712.2	4.94
P4-180-2	−669.2	−660.2	2.25	12%/0.5 h	−641.8	−647.8	1.58
P4-180-12	−715.1	−693.5	4.09	12%/7 h	−717.8	−702.2	4.36
P4-180-65	−718.3	−710.2	4.67	12%/48 h	−724.6	−711.6	4.86

3.3. Precipitate Features of As-Strained and Aged Specimens

The microstructural features of Al alloys with different pre-strains and aging treatments were further investigated using TEM. Figure 8 shows TEM bright-field images and corresponding selected-area electron diffraction (SAED) patterns of the 2A97 Al-Cu-Li alloy with the 0% pre-deformed alloy aged at 160 °C for different times. By detailed analysis of SAED patterns acquired along different zone axes, together with TEM images, the types of precipitates can be recognized, and their sizes measured using ImageJ software (version 1.54t, National Institutes of Health, Bethesda, MD, USA). For each condition, at least 100 precipitates were randomly selected from multiple TEM micrographs for statistical analysis, and the average values are reported in this work. As seen in Figure 8, with increasing aging time, the precipitate types remain essentially unchanged, mainly consisting of the T_1 , δ' , θ' , and S' phases. However, their size and number density change with the aging time. In addition, the characteristics of grain-boundary precipitates also evolve. As shown in Figure 8a, for the P0-160-22 (UA) alloy, the SAED pattern along the $[112]_{\text{Al}}$ zone axis shows streaks from the T_1 phase and weak spots from the δ' phase. The corresponding TEM image reveals several T_1 plates with an average size of ~100.5 nm, along with spherical δ' particles and a few S' plates. Figure 8b is a TEM image of grain boundaries (GBs), showing that the GB phases are finely distributed and continuous. Along the $[100]_{\text{Al}}$ zone axis, streaks of the θ' phase are visible (see Figure 8c), and the corresponding TEM image shows a dispersed distribution of θ' plates with an average size of about 88.3 nm. As shown in Figure 8d–f, for the P0-160-84 (PA) alloy, visible T_1 , δ' , and θ' phases are present, along with a small amount of S' . The GB phases are still continuously distributed. Compared with the UA state, the number density of δ' decreases, while its size increases slightly to about 38.6 nm. Both the number and size of T_1 and θ' phases increase moderately, reaching average sizes of ~129.7 nm and ~117.4 nm, respectively, for the P0-160-300 alloy (OA, see Figure 8g–i). The types of precipitates do not change, but the T_1 plates coarsen significantly in size, while the θ' plates decrease in both number and size. The average sizes of T_1 and θ' are ~168.6 nm and ~87.4 nm, respectively. It suggests that during over-aging, the θ' phase undergoes partial re-dissolution. The number density of δ' further decreases, but its average size increases to about 48.3 nm. The GB precipitates become coarse and discontinuous.

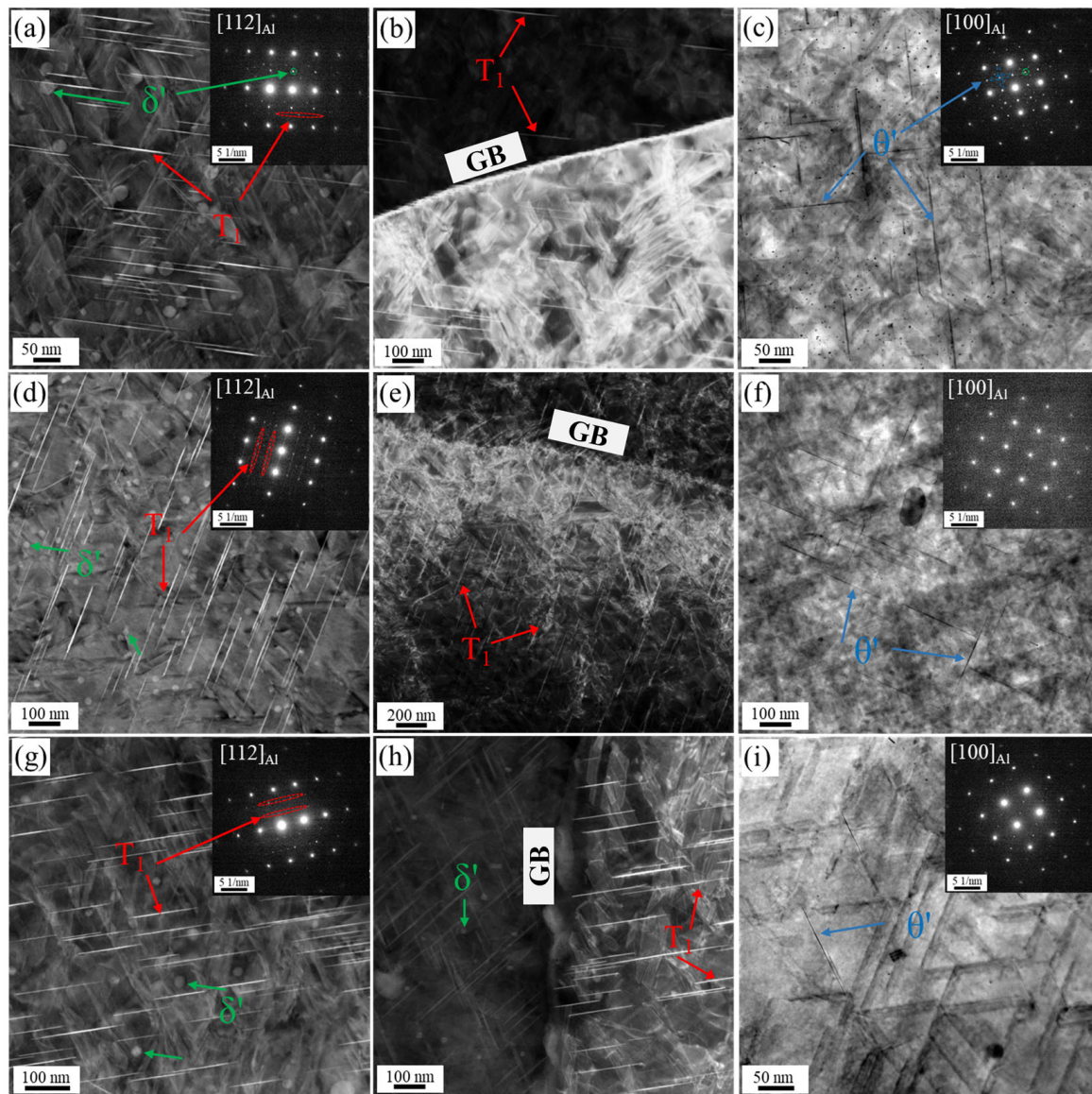


Figure 8. TEM images and SAED patterns of the 0% pre-deformed alloy aged at 160 °C for different times: (a–c) 22 h; (d–f) 84 h; (g–i) 300 h.

Figure 9 shows TEM images of the 4% pre-strained alloy aged at 160 °C for different times. Compared with the case displayed in Figure 8, the 4% pre-deformation does not change the types of precipitates. However, the precipitate sizes are significantly reduced. The evolution of precipitate size and number density with aging time follows the same trend as in the 0% pre-deformed alloy. As seen in Figure 9a, for the P4-160-6 (UA) alloy, the SAED pattern along $[112]_{Al}$ shows T_1 streaks and weak δ' spots. Many finely and uniformly distributed T_1 plates are seen with an average size of ~ 52.8 nm, along with spherical δ' (average size ~ 27.3 nm) and a few S' plates. Figure 9b shows that the GB phases are few, fine, and discontinuous. Figure 9c presents the SAED pattern along $[110]_{Al}$ and the corresponding TEM image, in which a few fine θ' plates are observed with an average size of ~ 41.3 nm. In the P4-160-6 condition (PA, see Figure 9d–f), abundant T_1 , δ' , and θ' phases are present, along with a small amount of S' . Compared with the UA state, the number and size of T_1 increase, reaching an average size of ~ 62.1 nm. The number of δ' decreases, but their size increases to ~ 31.4 nm. The θ' plates increase significantly, and their average size grows from ~ 41.3 nm to ~ 63.7 nm. Moreover, the GB phases form a continuous distribution due to local growth and coarsening.

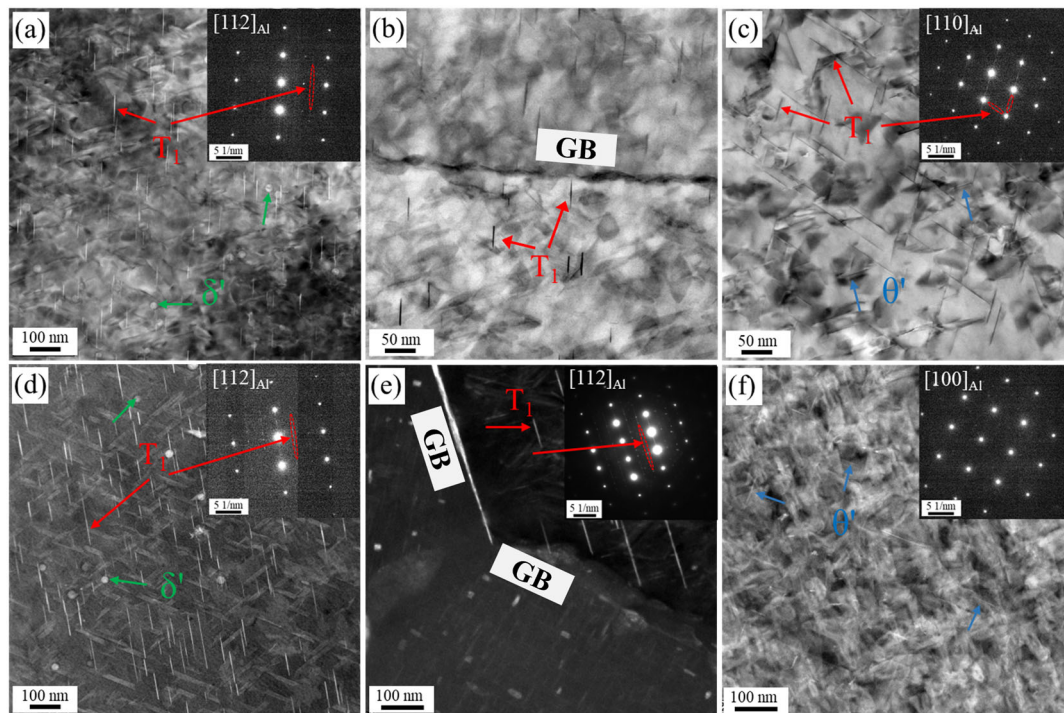


Figure 9. TEM images and SAED patterns of the 4% pre-deformed alloy aged at 160 °C for different times: (a–c) 6 h and (d–f) 60 h.

Figure 10 shows the typical TEM images of the alloy with 8% pre-deformation aged at 160 °C. The precipitate types remain the same. As shown in Figure 10a, for the P8-160-3 (UA) alloy, the SAED pattern along $[112]_{Al}$ and the corresponding TEM image show a large number of finely and uniformly distributed T_1 plates and spherical δ' particles, with average sizes of ~ 38.6 nm and ~ 24.2 nm, respectively, along with a few S' plates. Figure 10b reveals fine and discontinuous GB phases, and Figure 10c suggests a small number of fine θ' plates with an average size of ~ 30.6 nm. As for the P8-160-48 specimen (PA, see Figure 10d–f), many T_1 plates are present, along with a few δ' and S' precipitates. It is also found that the T_1 size increases with aging time, reaching an average of ~ 43.3 nm. The GB phases become continuously distributed and are shown in Figure 10g–i (P8-160-150); the T_1 plates exhibit obvious coarsening, with an average size of ~ 52.2 nm. Coarse and discontinuous GB phases are observed, which are attributed to the larger pre-strains.

Figure 11 shows TEM images of the alloy with 12% pre-deformation, aged at 160 °C. As seen in Figure 11a–c, corresponding to the P12-160-3 (UA) alloy, many finely and uniformly distributed T_1 plates (average size ~ 35.1 nm) and spherical δ' (~ 22.1 nm) are generated, along with a few S' plates. Figure 11b shows fine and discontinuous GB phases, which are not different from the observations in the above-mentioned specimens. As for the P12-160-38 (PA) alloy, lots of T_1 plates are present, along with small amounts of δ' , θ' , and S' (see Figure 11d–f). Notably, the average size of δ' decreases slightly to ~ 19.5 nm, probably due to aging-induced redissolution. The average sizes of T_1 and θ' in such peak-aged specimens are ~ 39.9 nm and ~ 34.2 nm, respectively. In addition, observed from Figure 11g–i, the T_1 plates are coarsened in the P12-160-100 specimen (OA), but their average size (~ 40.6 nm) does not increase dramatically. No δ' or θ' phases are observed. Coarse and discontinuous GB phases are present. In summary, at an aging temperature of 160 °C, increasing pre-deformation strain promotes the nucleation and growth of the T_1 phase, due to the increased dislocations and strain energy for precipitate formation. Significantly, the continuous growth of the preferentially nucleated T_1 plates commonly

suppresses the precipitation of the θ' phase; consequently, only a small amount of θ' is observed in the alloys with 8% and 12% pre-deformation. With prolonged aging, the T_1 plates grow and eventually coarsen significantly. The GB phases evolve with aging time from fine and discontinuous to fine and continuous, and finally to coarse and discontinuous, which contributes significantly to the deterioration of mechanical properties in the OA state.

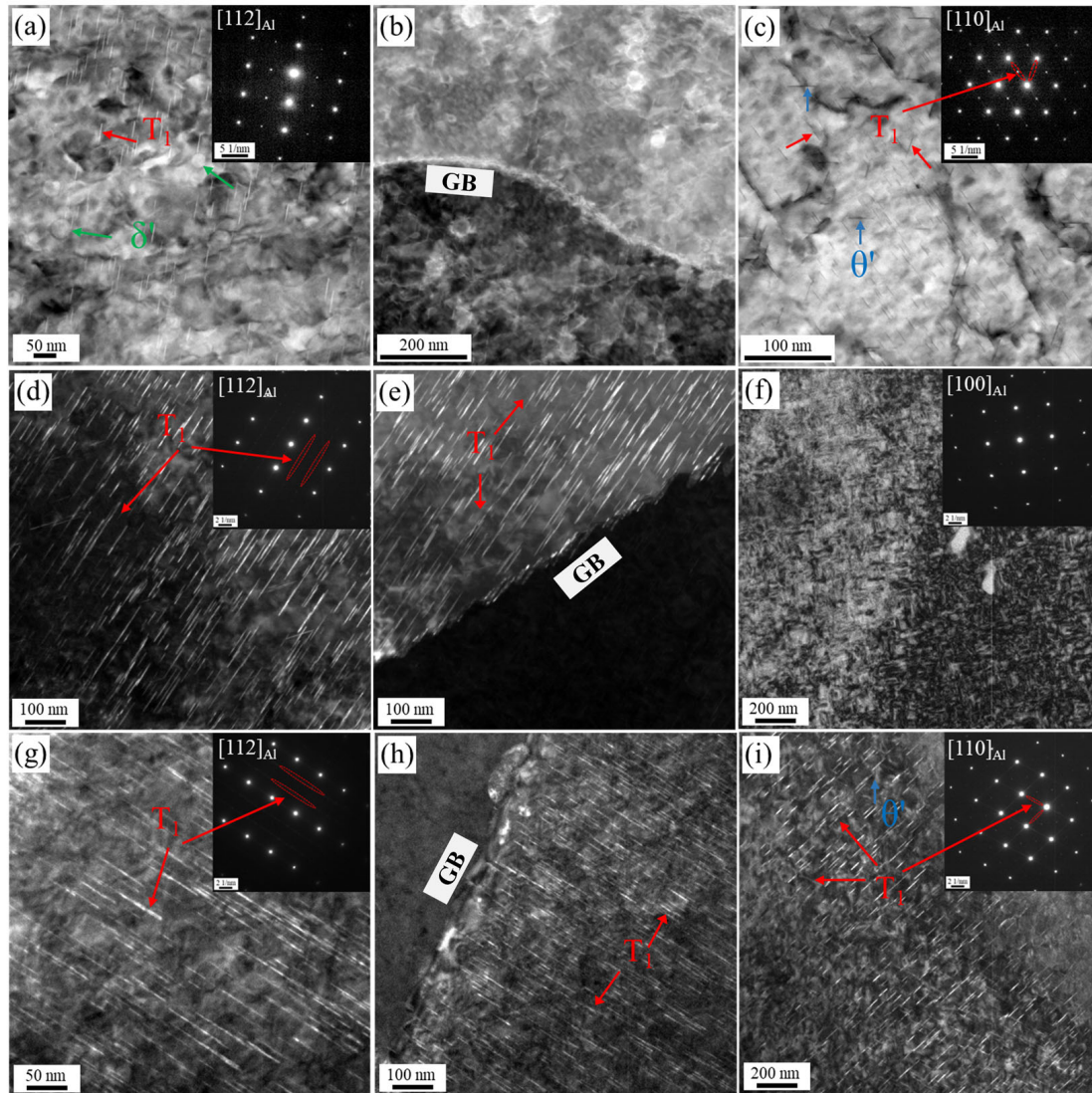


Figure 10. TEM images and SAED patterns of the 8% pre-deformed alloy aged at 160 °C for different times: (a–c) 3 h; (d–f) 48 h; (g–i) 150 h.

The precipitate morphologies of the 180-aged alloys were also examined. Figure 12 shows a series of TEM images of the alloys without pre-deformation aged at 180 °C for different times. Compared with aging at 160 °C, raising the temperature to 180 °C does not change the precipitate types, but the precipitate sizes increase significantly. As seen in Figure 12a–c (corresponding to the P0-180-3 alloy, UA), many finely distributed T_1 plates (average size ~92.2 nm), along with spherical δ' (~30.3 nm) and a few S' plates, are found, as confirmed by the SAED pattern along $[112]_{Al}$. Moreover, GB phases are fine and discontinuous (see Figure 12b). A small number of fine θ' plates (~60.6 nm) are also observed (see Figure 12c). As shown in Figure 12d–f for the P0-180-36 specimen (PA), abundant T_1 , δ' , and θ' phases are present, with a small amount of S' . Compared with the UA state, T_1 plates increase to ~127.8 nm; δ' decreases in number but grows to ~36.9 nm; and θ' significantly increases in both number and size (~110.7 nm). The GB phases become

continuously distributed. For the P0-180-85 specimen (see Figure 12g–i, OA), T_1 plates coarsen further (~ 141.1 nm), while θ' decreases markedly, indicating partial re-dissolution of θ' and δ' . The GB phases become coarse and discontinuous. From the above observation, it is seen that with increasing aging time, the sizes of T_1 and δ' gradually increase, while θ' first increases and then decreases.

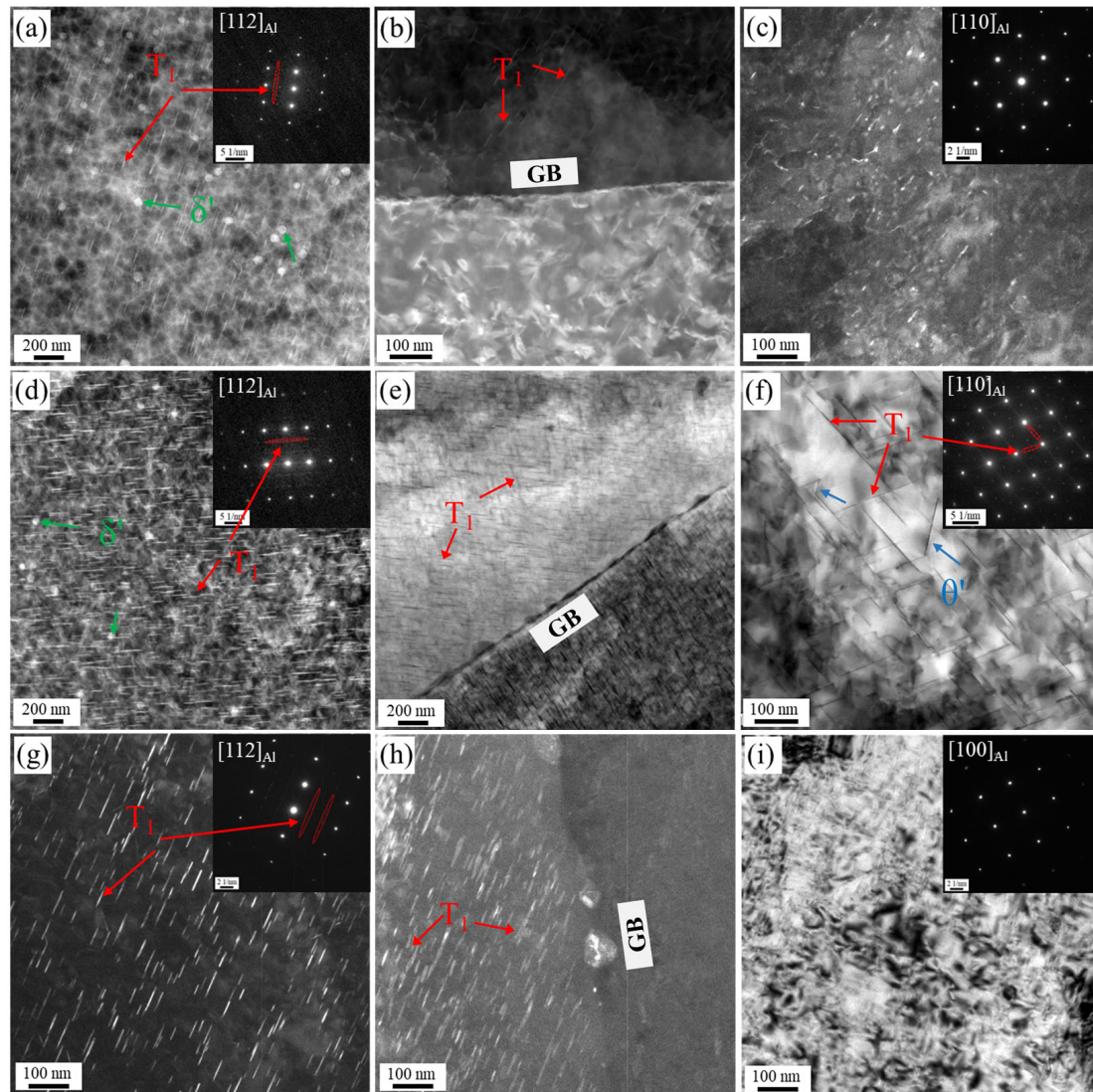


Figure 11. TEM images and SAED patterns of the 12% pre-deformed alloy aged at 160 °C for different times: (a–c) 3 h; (d–f) 38 h; (g–i) 100 h.

Figure 13 presents the microstructure of the alloy with 4% pre-deformation, aged at 180 °C, as seen in Figure 13a–c, fine and uniformly distributed T_1 plates (~ 52.4 nm) form in the P4-180-3 specimen (UA), as confirmed by the SAED pattern, along with minor δ' and S' phases. GB phases are fine and discontinuous in this case. As the aging time is extended, more T_1 plates are present in the P4-180-12 specimen (see Figure 13d–f, PA), along with minor δ' , θ' , and S' phases. The newly generated θ' plates increase notably in number, reaching an average size of ~ 82.9 nm. The GB phases become continuously distributed in such a peak-aged alloy. Moreover, for the P4-180-65 (OA) alloy, δ' and θ' are nearly absent, while T_1 plates coarsen to ~ 74.6 nm (see Figure 13g–i). Notably, sub-GBs are decorated with fine, continuous T_1 plates, as displayed in Figure 13h, whereas high-angle GBs have coarse and discontinuous phases.

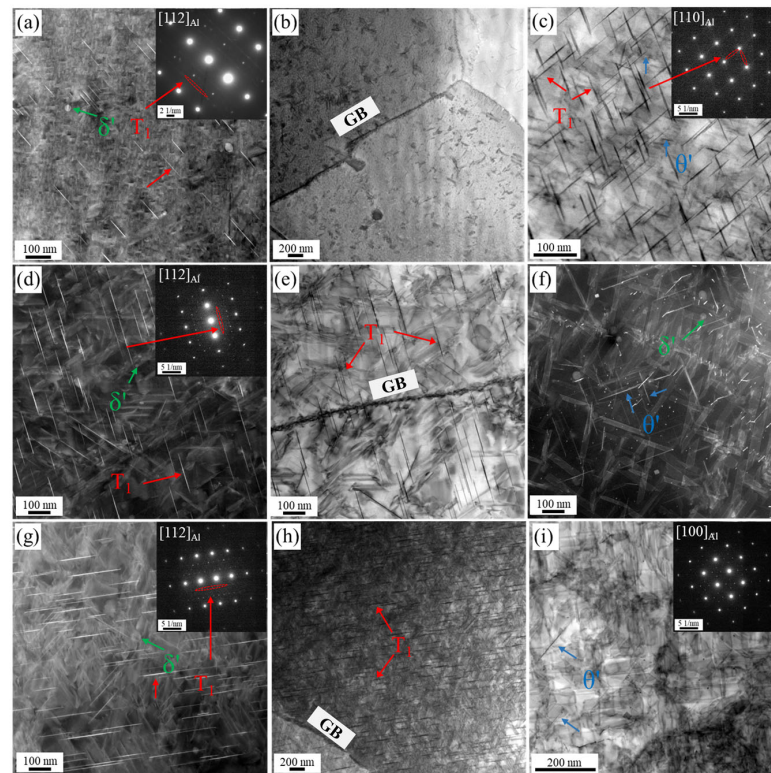


Figure 12. TEM images and SAED patterns of the 0% pre-deformed alloy aged at 180 °C for different times: (a–c) 3 h; (d–f) 36 h; (g–i) 85 h.

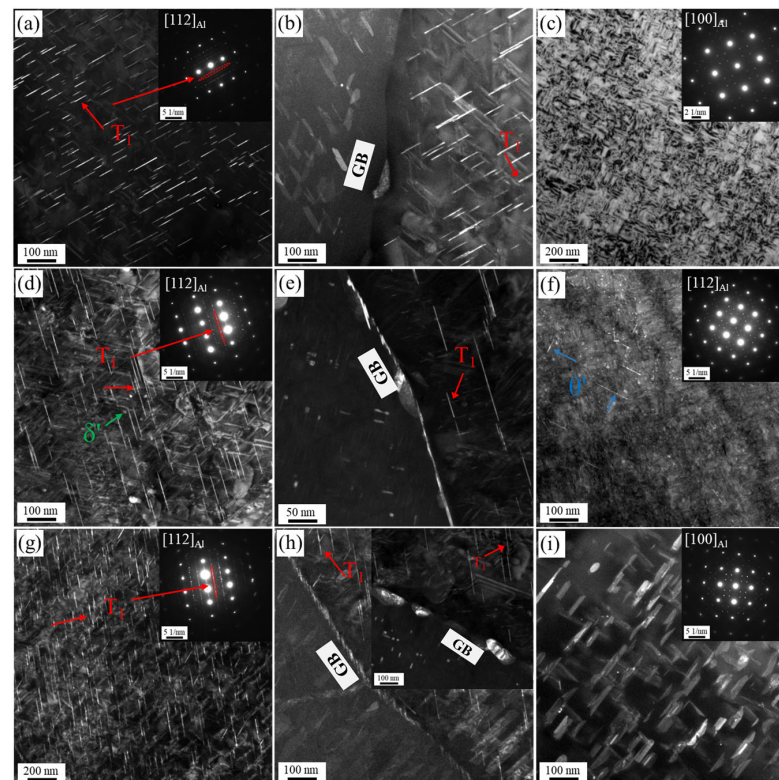


Figure 13. TEM images and SAED patterns of the 4% pre-deformed alloy aged at 180 °C for different times: (a–c) 3 h; (d–f) 12 h; (g–i) 65 h.

Figure 14 shows the typical microstructure of the 8% pre-deformed alloy aged at 180 °C; no θ' phase is observed, indicating that 8% pre-strain combined with aging at 180 °C suppresses θ' precipitation. As shown in Figure 14a–c, the P8-180-1.5 specimen

(UA) exhibits the presence of fine and uniformly distributed T_1 plates (~ 45.4 nm) with minor S' , as confirmed by the strong T_1 streaks in the $[112]_{Al}$ SAED pattern. Notably, the $[100]_{Al}$ SAED pattern (see Figure 15c) shows no θ' streaks or spots, confirming the absence of θ' . As for the P8-180-10 (PA, see Figure 14d–f) and P8-180-48 specimens (OA, see Figure 14g–i), it is found that the T_1 plates increase in number and size with aging time, especially coarsening in the over-aged stage. GB phases evolve from fine and discontinuous (UA state) to coarse and continuous (PA state), and finally to discontinuous (OA state).

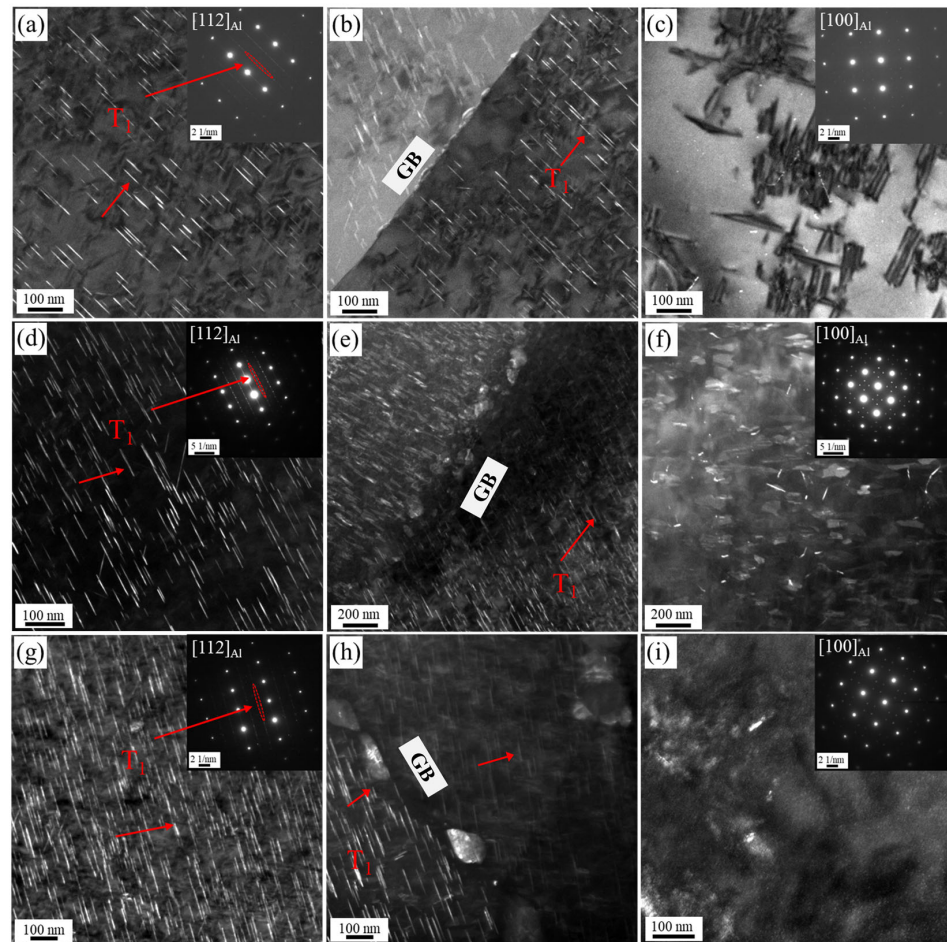


Figure 14. TEM images and SAED patterns of the 8% pre-deformed alloy aged at 180 °C for different times: (a–c) 1.5 h; (d–f) 10 h; (g–i) 48 h.

Figure 15 displays the microstructure of the 12% pre-strained alloy aged at 180 °C. Similar to the 8% condition, the precipitates are mainly T_1 , δ' , and S' ; θ' is absent, further confirming that high-temperature aging (180 °C) combined with large pre-deformation (8% or 12%) is unfavorable for θ' precipitation. The TEM image in Figure 15a–c, regarding the UA state, reveals uniformly distributed T_1 plates (~ 40.2 nm) and a small amount of S' phase. Fine GB phases are discontinuous. Moreover, as shown in Figure 15d–f (for the P12-180-7, PA), considerable T_1 plates are present, along with minor δ' . The average T_1 size increases to ~ 51.9 nm, and GB phases become continuously distributed. With aging, the OA regime increases to 48 h (see Figure 15f), and the average T_1 size decreases to ~ 46.8 nm (a reduction of ~ 12.1 nm compared to PA). It is found in Figure 15g–i that there are many short, rugby-shaped T_1 plates and elongated, tadpole-shaped T_1 plates, suggesting that prolonged aging at 180 °C with 12% pre-deformation leads to dissolution of the T_1 phase. In summary, compared with 160 °C, the types of precipitates in these 180 aged specimens

change. In particular, the θ' phase completely disappears, and the other precipitates exhibit larger sizes when aged at 180 °C.

From the above TEM observations, the following general trends of precipitate evolution with pre-straining and post-aging treatment parameters are evident. At the same pre-deformation and aging temperatures, the average precipitate size generally increases with increasing aging time, as is widely reported for many Al alloys. Specifically, the δ' phase coarsens while its number density gradually decreases; the T_1 phase also coarsens, with its number density first increasing and then stabilizing; the θ' phase shows a gradual decrease in number density, and its size first increases and then decreases. An exception is observed for the 180 °C-aged alloy with 12% pre-strain, where prolonged aging results in a reduction in T_1 size, suggesting partial dissolution. At the same aging temperature of 160 °C, increasing the pre-deformation strain significantly reduces precipitate size. However, in the alloys with 8% and 12% pre-strain, the extensive precipitation of T_1 suppresses the precipitation of θ' and δ' . When the aging temperature is raised to 180 °C and the pre-strain is increased to 8% or 12%, the precipitation of θ' becomes unfavorable. Under identical pre-deformation and aging times, a higher aging temperature promotes precipitate coarsening. Furthermore, with prolonged aging, the GB phases generally evolve from discontinuous to continuous and then back to discontinuous.

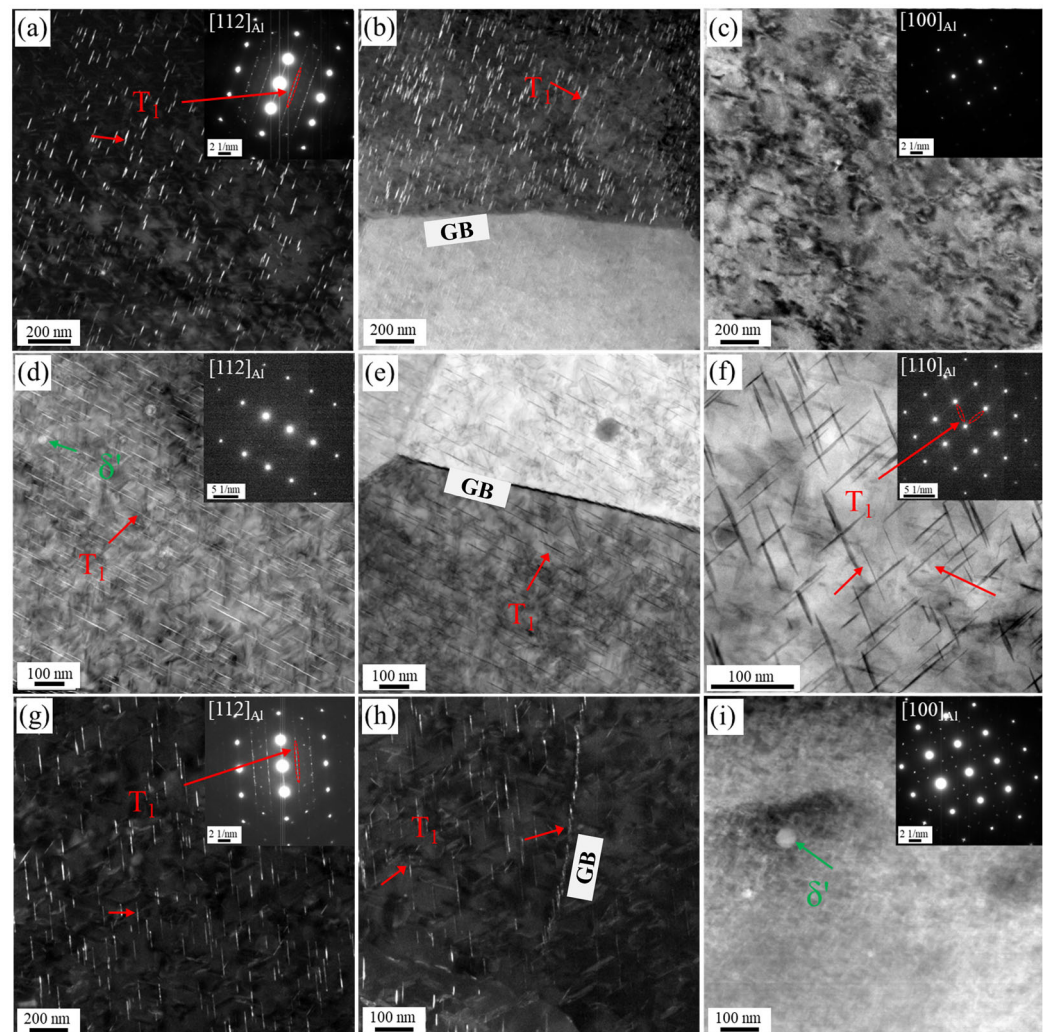


Figure 15. TEM images and SAED patterns of the 12% pre-deformed alloy aged at 180 °C for different times: (a–c) 0.5 h; (d–f) 7 h; (g–i) 48 h.

4. Discussions

4.1. Influences of Aging Condition on Precipitation Behavior

As widely reported, the common precipitate phases in Al-Cu-Li-based alloys include T_1 (e.g., Al_2CuLi), δ' (e.g., Al_3Zr in those with certain rare-element alloying), θ' (e.g., Al_2Cu), θ'' /GP-II zones (e.g., Al_3Cu), GP-I zones, and S' (Al_2CuMg) [34–36]. The types, probabilities, and sequence of precipitation are primarily determined by the alloy's composition. Especially, it is well recognized that the Cu/Li ratio has a significant influence on the precipitate types and precipitation sequence in Al-Cu-Li alloys [11]. The present 2A97 Al-Cu-Li alloy sheet used in this study has a Cu/Li ratio of approximately 2.47, so the precipitation sequence should fall into the type in which the Cu/Li ratio is 1–2.5, namely α (SS) $\rightarrow$ GP zones + δ' $\rightarrow$ $\delta' + \theta'$ $\rightarrow$ $\delta' + T_1$ $\rightarrow$ T_1 . As shown in the TEM images above, the main precipitate phases observed in this work under different aging conditions include T_1 , δ' , θ' , and S' . At 160 °C, increasing the pre-deformation strain from 0% to 12% does not change the types of precipitates. With increasing aging time, the number density of δ' decreases while its size increases. In the early aging stage, the supersaturated solid solution contains a large number of vacancies, and δ' (a metastable phase coherent with the matrix) has a low nucleation energy, leading to its homogeneous and extensive precipitation. At the peak-aged and over-aged stages, δ' grows slowly. Meanwhile, the growth of abundant T_1 plates causes the dissolution of small δ' particles, leaving only a few large δ' precipitates. The number density of θ' shows a similar trend with aging time, but its size first increases and then decreases.

During the precipitation sequence of heat-treatable Al-Cu-Li alloys, GP zones (solute clusters) fully coherent with the matrix and having the lowest interfacial energy form first [37]. Hence, the number of θ' precipitates would be expected to increase gradually. However, GP zones exist mainly in the early aging stage, so the transition from GP zones + δ' to $\delta' + \theta'$ occurs during the under-aged stage. This is consistent with our observation that the number of θ' precipitates at the peak-aged condition does not increase compared to the under-aged condition. The T_1 phase, a semi-coherent equilibrium phase with the matrix, precipitates abundantly throughout the entire aging process. Raising the aging temperature promotes solute diffusion, provides more energy fluctuations for nucleation, and thus accelerates the nucleation and growth of precipitates [38]. The coarsening of grain-boundary phases is particularly pronounced in the under-aged stage. Moreover, the extensive precipitation of T_1 consumes small θ' and δ' phases. As aging proceeds, T_1 plates grow continuously, competing for Li and Cu atoms, leading to the dissolution of θ' and δ' .

Significantly, introducing pre-deformation generates a high density of defects, like vacancies and dislocations. Dislocations provide fast diffusion paths for solute atoms, reducing the activation energy for diffusion. They also serve as additional nucleation sites for the T_1 phase. Furthermore, dislocations possess high energy that can facilitate the nucleation of T_1 [39]. Consequently, pre-deformation significantly promotes the precipitation of many fine and uniformly distributed T_1 plates. Compared with the T6 (0% pre-deformation) condition, the pre-deformed alloys exhibit much finer T_1 plates and a higher number density. The extensive precipitation of T_1 consumes θ' and δ' , leading to a reduction in both their number density and size with increasing pre-deformation. In the alloy with 12% pre-deformation aged at 180 °C, partial re-dissolution of T_1 is observed in the over-aged stage. At this point, the small amount of δ' remaining cannot provide sufficient Li atoms for T_1 growth. In addition, the excessively high dislocation density introduced by large pre-deformation may cause localized stress concentration. Therefore, under the combined conditions of high pre-deformation and high-temperature over-aging, fine T_1 plates dissolve.

Throughout the aging process, the GB phases are mainly Al-Cu and Al-Cu-Li phases. Their distribution generally evolves from discontinuous to continuous and then back to discontinuous. In the under-aged stage, solute atoms and vacancies have not yet sufficiently diffused to the grain boundaries; the grain boundaries are deficient in solutes, resulting in a low precipitation driving force and the formation of a few fine, discontinuous phases. In the peak-aged stage, grain boundaries, as high-energy regions, provide abundant nucleation sites for solute atoms that have adequately diffused to the boundaries, leading to a continuous distribution of GB phases. As aging extends into the over-aged stage, the continuously growing and coarsening T_1 plates demand more Li and Cu atoms, causing a reduction in small precipitates at the grain boundaries. Due to solute atom redistribution, discontinuous precipitates appear at the GBs.

4.2. Influence of the Aging Condition on Mechanical Properties

As described in Section 3.3, the T_1 , δ' , θ' , and S' phases precipitate to varying extents in the 2A97 Al-Li alloy under different aging conditions. The uniform distribution of these precipitates significantly pins dislocations and impedes their movement, thereby enhancing the alloy's strength [40,41]. For the specimens aged at 160 °C, with increasing aging time, the strength first increases, then remains stable, and finally decreases, while the elongation shows an overall declining trend. In the early aging stage, namely the under-aged condition, fine T_1 and θ' plates precipitate within the grains. T_1 and θ' are semi-coherent precipitates in Al-Cu-Li alloys, and the precipitation strengthening mechanism is primarily the Orowan bypassing mechanism [42,43]. When these precipitates are small, the shear stress required for dislocations to bypass them is low, resulting in weak obstruction of dislocations. Consequently, the under-aged alloy exhibits low strength but good elongation. With prolonged aging, the θ' plates decrease, while the number and size of T_1 plates increase. In Al-Cu-Li alloys, T_1 and θ' are usually the main strengthening precipitates, and the strengthening effect of T_1 is greater than that of θ' .

Meanwhile, T_1 is also the main brittle phase in Al-Cu-Li alloys [44]. In the peak-aged condition, the number density and size of T_1 plates increase, leading to higher strength but lower elongation. After peak aging, the number of T_1 plates does not change significantly, but their size continues to increase, and some coarsening occurs. Coarsened T_1 plates contribute less to strength and increase the brittleness of the alloys. Moreover, in the over-aged condition, the distribution of coarse T_1 plates becomes non-uniform. As a result, both the strength and elongation decrease in the over-aged state. Moreover, it is well known that raising the aging temperature accelerates precipitation and precipitate growth, thereby increasing the age-hardening rate and shortening the time required to reach the peak-aged state [45]. When the aging temperature is increased to 180 °C, the T_1 plates become larger, and the θ' plates increase in size but decrease in number. The alloy is then dominated by relatively large T_1 plates. Consequently, the strength and hardness increase, but elongation decreases significantly. These results indicate that the evolution of the mechanical properties is closely associated with the precipitation behavior during aging. As the strengthening precipitates continuously precipitate, grow, and coarsen, their size and distribution are gradually modified, resulting in the corresponding evolution of the mechanical properties. This interpretation is consistent with previous studies [46,47] on Al-Cu-Li alloys.

Furthermore, pre-deformation significantly influences the precipitation behavior by refining the precipitates and making their distribution more uniform [48,49]. Fine, uniformly distributed precipitates disperse slip via the bypass mechanism, reducing stress concentrations and thus enabling the alloy to achieve high strength while maintaining good elongation. Therefore, an appropriate level of pre-deformation is an effective approach to improve the comprehensive mechanical performance of the alloy by optimizing the

precipitation characteristics. However, different amounts of pre-deformation have different effects on the properties. During aging at 160 °C, when the pre-deformation increases from 0% to 4%, the sizes of T_1 and θ' in the peak-aged alloy decrease significantly.

In contrast, the number of T_1 plates increases markedly, resulting in improved strength while retaining relatively good elongation. As the pre-deformation increases from 4% to 8%, the number of nucleation sites for T_1 increases, leading to a higher T_1 density and, consequently, an increase in strength. At a pre-deformation of 12%, the alloy exhibits high strength but very low elongation. A continuous increase in pre-deformation leads to a high dislocation density. The interactions among dislocations and the obstruction of dislocations by T_1 plates can readily lead to localized stress concentrations.

Additionally, the irregular edges of T_1 plates can become sites for microcrack initiation. Hence, the plasticity decreases, and the elongation reduces. This indicates that a larger pre-deformation is not always better. The present study shows that the alloy with 4% pre-deformation exhibits the best combination of mechanical properties. Although increasing the pre-deformation beyond 4% further accelerates precipitation and slightly increases the strength, the improvement becomes marginal, whereas the loss of elongation becomes more pronounced. Therefore, an appropriate pre-deformation level, rather than the maximum pre-deformation, is more beneficial for optimizing the mechanical performance of the 2A97 Al-Cu-Li alloy. We should also note that phases forming at GBs affect the alloy's plasticity. In the under-aged stage, fine, discontinuous GB phases result in good deformation uniformity during fracture and thus higher elongation. In the peak-aged stage, the number of GB phases increases and becomes continuously distributed, which readily induces intergranular fracture and reduces elongation. In the over-aged stage, such phases are coarse and discontinuous. Coarse GB phases easily cause stress concentration, and their irregular boundaries can promote crack initiation, leading to the poorest elongation. In addition to influencing the tensile properties, the evolution of grain-boundary precipitates also plays an important role in determining the corrosion behavior of the alloy. The relationship between grain-boundary precipitation and corrosion resistance is discussed in the following section.

4.3. Influences of Aging Conditions on Corrosion Resistance

The microstructure of the present Al-Cu-Li alloys, particularly the intragranular precipitates, GB phases, and precipitate-free zones (PFZs), strongly influences corrosion performance. Grain boundaries are preferential sites for solute segregation, precipitation, and localized corrosion. Their structural characteristics can influence grain-boundary diffusion, microstructural evolution, and corrosion susceptibility, thereby affecting the overall corrosion performance of aluminum alloys [50,51]. As described in Section 3.3, differences in the type, size, number, and distribution of intragranular precipitates, as well as the morphologies of GB phases, generated under different aging conditions, should lead to variations in IGC and electrochemical corrosion behavior. As shown in Figure 4a, the P0-160-22 (UA) alloy exhibits typical localized intergranular corrosion. The corresponding TEM image (see Figure 8) reveals a PFZ of a certain width along the GBs, with distinct T_1 plates at the boundaries. The interaction between the T_1 plates at GBs and the adjacent PFZ leads to anodic dissolution of T_1 , allowing intergranular corrosion to propagate along the grain boundaries. This interpretation is consistent with the IGC morphologies shown in Figures 4 and 5. Under the under-aged condition, localized intergranular corrosion is dominant because the corrosion path is mainly confined to the PFZ adjacent to the grain boundaries. With increasing aging time, the evolution of grain-boundary precipitates and PFZs changes the corrosion morphology, resulting in a gradual transition from localized intergranular corrosion to pitting corrosion. In this process, the PFZ provides a corrosion path for the propagation of intergranular corrosion.

Meanwhile, the relatively small number of T_1 plates inside the grains makes the intragranular potential less negative, resulting in a large potential difference between the grain interior and the grain boundaries, which drives corrosion. Previous work [52] has shown that the OCP primarily reflects the potential of the grain interior, as the GB area fraction is very small. Therefore, the intragranular potential can be more intuitively reflected by measuring the OCP. With increasing aging time, the number and size of T_1 plates increase, the OCP shifts negatively, the potential difference between the grain interior and grain boundaries decreases, and the driving force for intergranular corrosion decreases. Concurrently, the PFZ gradually disappears, eliminating the corrosion path. As a result, the aged alloy's susceptibility to intergranular corrosion decreases. As the aging temperature increases, precipitate size increases. In the under-aged stage, the GB phases coarsen more significantly, leading to a greater potential difference between the grain interior and boundaries, a higher corrosion-driving force, and thus increased susceptibility to intergranular corrosion.

Figure 16 shows the distribution map of electrochemical OCP values and intergranular corrosion for the alloy at different aging times. Two main regions can be distinguished. The intergranular corrosion region corresponds to more positive OCP values, while the pitting region corresponds to more negative OCP values. When the OCP is more positive, the potential difference between the grain interior and boundaries is larger, providing a greater driving force for intergranular corrosion. As pre-deformation increases, a large number of fine T_1 plates precipitate uniformly within the grains. The OCP shifts negatively, reducing the potential difference between the grain interior and boundaries and thus diminishing the driving force for intergranular corrosion. At the same time, the increased pre-deformation eliminates the PFZ along the GBs, blocking the corrosion path for intergranular corrosion. Consequently, the susceptibility to intergranular corrosion decreases, and varying degrees of pitting occur instead. Pitting is primarily caused by intragranular corrosion. With increasing pre-strain, the number and size of intragranular T_1 plates gradually increase, the intragranular potential continues to shift negatively, the potential difference between the grain interior and boundaries becomes smaller, and the driving force for intergranular corrosion is further reduced. Initially, the more negative T_1 plates near the grain boundaries undergo cyclic anodic dissolution with the surrounding Al matrix. Because the potential difference between the grain interior and boundaries is small, corrosion does not always propagate along the grain boundaries; instead, it spreads gradually into the grains, leading to extensive pitting.

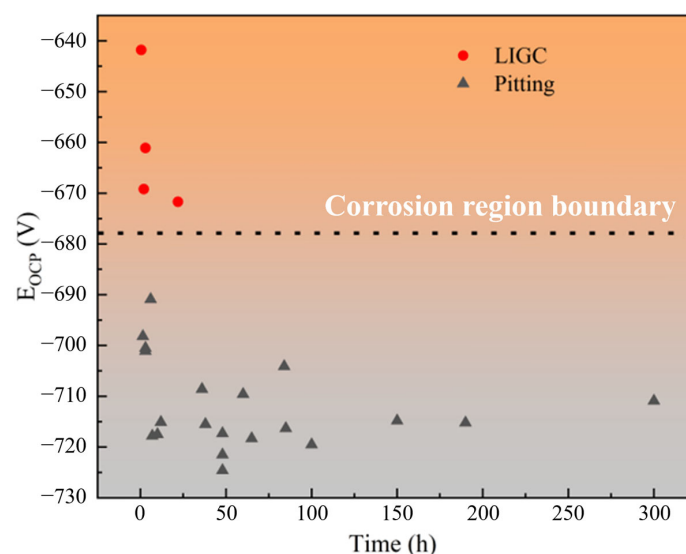


Figure 16. Distribution map of electrochemical OCP values and intergranular corrosion of the present 2A97 Al-Cu-Li alloys at different aging times.

5. Conclusions

This work systematically investigated the effects of pre-deformation strains (including 0%, 4%, 8%, 12%), aging temperature (i.e., 160 °C and 180 °C), and aging time (from 1 h to 200 h) on the mechanical properties, corrosion behavior, and microstructure of a classic 2A97 Al-Cu-Li alloy. The main conclusions are displayed as follows:

- (1) The main precipitates (i.e., T_1 , δ' , θ' , S') remain consistent across aging conditions. With longer aging time, T_1 coarsens, δ' decreases in number but increases in size, and θ' decreases after an initial increase in size. Raising the aging temperature to 180 °C increases precipitate size but does not alter the evolution trend, except for the θ' phase.
- (2) Increasing pre-deformation strain from 0% to 12% at 160 °C increases T_1 number density and reduces its size. In the 180 °C-aged specimens, θ' disappears in 8% and 12% pre-deformed alloys, and T_1 partially re-dissolves in the over-aged 12% pre-deformed alloy.
- (3) The morphologies of grain-boundary phases evolve from fine/discontinuous to fine/continuous, and finally to coarse/discontinuous with aging. Pre-deformation refines such GB precipitates, whereas a higher aging temperature promotes coarsening.
- (4) Higher aging temperature and larger pre-deformation strain accelerate precipitation kinetics and shorten peak-aging time. The alloy strength increases, but the total elongation decreases with greater pre-deformation or higher aging temperature. The 4% pre-deformed alloy aged at 160 °C offers the best combination of strength and ductility.
- (5) Intergranular corrosion susceptibility increases with aging temperature, mainly due to coarsened grain-boundary phases. With prolonged aging, the increased precipitation and coarsening of intragranular T_1 reduce the electrochemical potential difference and promote the transition from localized intergranular corrosion (LIGC) to pitting, thereby improving the intergranular corrosion resistance. Increasing pre-deformation from 0% to 12% increases intragranular T_1 precipitation, reduces the potential difference, and therefore lowers intergranular corrosion susceptibility.

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