

Article

The Influence of Different Aging Temperatures on the Microstructure and Corrosion Behavior Evolution Characteristics of the Al-Cu-Li Alloy

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Abstract

In the current work, the microstructural characteristics and corrosion performance of an Al-3.6Cu-1.0Li-0.40Mg-0.32Mn-0.12Zr alloy are correlated across different artificial aging regimes (150 °C, 160 °C, and 170 °C). In the under-aging stage, the corrosion depth increases with rising aging temperature, from 342.86 µm at 150 °C to 495.13 µm at 170 °C, indicating deteriorated corrosion resistance at higher temperatures. This trend is closely related to the significant increase in the proportion of the T1 phase in the matrix's primary precipitate. Upon artificial aging for 24 h, the hardness increases gradually as the aging temperature rises. At higher aging temperatures, short-term aging hardness is higher, likely due to the formation of the T1 phase, which can also provide a strengthening effect. In contrast, the corrosion resistance of the alloy is enhanced at higher aging temperatures after 24 h of aging. These corrosion phenomena are closely related to the dominance of the θ'' phase during low-temperature aging and the gradual increase in the S' phase during high-temperature aging. Furthermore, a transition from intergranular corrosion to pitting corrosion is identified at the high aging temperature of 170 °C with extended aging time. This corrosion mode transformation behavior is speculated to result from intermittent formation of magnesium segregation near the grain boundary, which alters the electrochemical heterogeneity between grain boundaries and the alloy matrix.

Keywords: Al-Cu-Li alloy; aging treatment; aging precipitation; corrosion behavior

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1. Introduction

The Al-Cu-Li alloy is a new, lightweight, high-strength alloy with many advantages, including high specific strength and stiffness, good corrosion resistance, and good fatigue resistance [1–3]. It is expected to be widely used in the aerospace and navigation fields. At present, Al-Cu-Li alloys are primarily used in the aerospace industry, particularly in key structural components such as fuel storage tanks [4,5]. However, the aviation industry's expectations for Al-Cu-Li alloys have not diminished, as attention to their multiple properties, including damage tolerance, corrosion resistance, and other attributes, has increased. Therefore, to move Al-Cu-Li alloys from the aerospace field to the aviation field,

to expand their applications in the maritime field with higher corrosion requirements in the future, and to achieve a multi-field, multi-purpose application pattern, detailed research and regulatory mechanisms for the corrosion performance of Al-Cu-Li alloys are crucial [6]. The currently proposed conceptual fourth-generation Al-Cu-Li alloy aims to improve multiple performance indicators, including comprehensive corrosion resistance.

Current regulations governing the corrosion performance of Al-Cu-Li alloys primarily focus on processing and heat-treatment technologies [6–8]. Usually, the precipitation phase is controlled to regulate the grain structure. In addition, many studies have focused on improving the corrosion performance of Al-Cu-Li alloys by modifying the internal precipitation potential and the electrochemical degree after precipitation [9–11]. This microstructure-corrosion regulation strategy has also been widely verified in other multi-component metallic material systems, providing a general research framework for corrosion performance optimization [12]. However, due to the complex, multi-precipitation system in Al-Cu-Li alloys, controlling the precipitation structure is indispensable for achieving key performance. Of particular concern is the complex precipitation behavior in Al-Cu-Li alloys [13]. Multiple research reports have shown that the addition of microalloying elements affects precipitation behavior without altering the types of precipitation phases. Previous studies have focused on the effects of adding certain elements or altering their content in Al-Cu-Li alloys, including magnesium, zinc, and silver [14–16]. Some elements have significantly improved the corrosion resistance of Al-Cu-Li alloys, while others have shown little improvement. In addition, many studies have focused on the influence of fine-grain regulation on grain structure, and, based on these findings, the effects of grain-structure regulation on corrosion resistance have been reported, suggesting the important role of an equiaxed fine-grain structure in improving corrosion resistance [17–19]. There are also many reports on the impact of aging systems on the corrosion resistance of Al-Cu-Li alloys. Numerous studies indicate that the corrosion resistance of the aging Al-Cu-Li alloy is significantly better than that of the T6-state Al-Cu-Li alloy [11]. The corrosion resistance of the Al-Cu-Li alloy treated by natural aging is significantly lower than that of the alloy treated by artificial aging. Compared with the ultra-high-strength 7-series aluminum alloy, some researchers have also proposed improving the corrosion resistance of Al-Cu-Li alloys through over-aging [20,21]. However, this approach did not yield consistent results across all Al-Cu-Li alloys with different compositions. This indicates that the complex precipitation system and behavior in Al-Cu-Li alloys will have a crucial impact on their corrosion performance [22–24]. Nevertheless, most existing studies focus on a single aging temperature or isolated aging states, and systematic comparison of corrosion evolution across the full aging cycle under multiple thermal regimes remains insufficient. In particular, the micro-mechanism of corrosion mode transition at elevated aging temperatures, as well as the regulatory role of grain-boundary Mg segregation and S' phase evolution, still lacks detailed clarification. To address these gaps, this work systematically investigates the precipitation behavior and corrosion performance of the Al-Cu-Li alloy at three aging temperatures.

The current research focuses on the precise correlation between aging precipitation behavior and corrosion performance of Al-Cu-Li alloys during artificial aging at 150 °C, 160 °C, and 170 °C. On one hand, this work systematically delineates the corrosion behavior evolution across the full-aging process at multiple temperatures, filling the research gap that most existing studies focus on isolated aging states without systematic temperature comparison. More importantly, this work reveals the temperature-dependent mechanism of the intergranular-to-pitting corrosion transition at elevated aging temperatures and clarifies the regulatory role of grain-boundary Mg segregation and S' phase evolution in this transition. This will provide robust support for basic corrosion data, providing sys-

tematic experimental data and theoretical support for the aging process optimization and corrosion performance regulation of Al-Cu-Li alloys.

2. Materials and Experimental Details

The materials presented in this paper were prepared by melting and casting in an argon atmosphere. The chemical composition of the alloy under study was measured using inductively coupled plasma optical emission spectrometry (ICP-OES), as shown in Table 1. The alloy ingot produced by the melting–casting method requires homogenization, hot rolling, and cold rolling to produce an aluminum–lithium alloy sheet approximately 2 mm thick [25]. The process flow adopted in the paper is shown in Figure 1. The sample alloy was heat-treated in a two-stage system: 460 °C for 8 h, followed by 540 °C for 24 h. Before hot rolling, the samples were held in an air furnace at 470 °C for 4 h to ensure uniform temperature. During hot rolling, the temperature was 480 °C and the total reduction rate was 80%. The samples were then annealed at 450 °C for 2 h and air-cooled to room temperature. The samples were then cold-rolled at room temperature to a total reduction of 60%. The cold-rolled sheet samples were then solution-treated at 505 °C for 50 min. The samples were quenched in water and then placed in a drying oven at 150 °C, 160 °C, or 170 °C for T6 artificial aging heat treatment for varying durations. These three types of alloy samples were labeled Sample-150, Sample-160, and Sample-170, respectively.

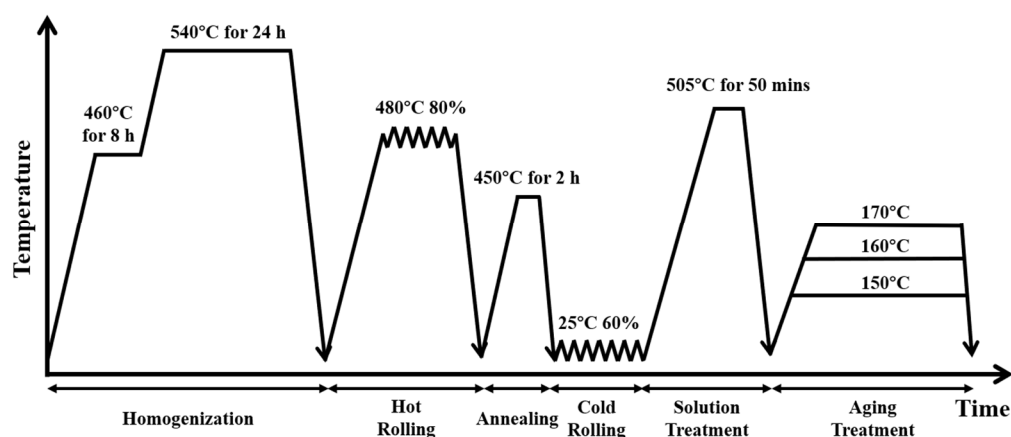


Figure 1. Schematic diagram of the heat treatment process for the studied Al-3.6Cu-1.0Li-0.40Mg-0.32Mn-0.12Zr alloys.

Table 1. Measured chemical composition of the three studied Al-3.6Cu-1.0Li-0.40Mg-0.32Mn-0.12Zr alloys, determined by ICP-OES (wt.%).

Cu	Li	Mg	Mn	Zr	Al
3.82	1.04	0.42	0.32	0.12	Bal.

The hardness of the prepared samples is then assessed using hardness testing. The hardness test is performed on a Vickers hardness machine (Laizhou Huayin Test Instrument Co., Ltd., Laizhou, China). The loading force during the test is 20 N, and the loading cycle duration is 15 s. Prior to testing, the specimens are ground with diamond abrasive paper in four successive grades of 200-grit, 400-grit, 800-grit, and 1200-grit. At the same aging time point, the hardness values for this heat-treatment state are obtained by averaging five replicate measurements and calculating error bars. The intergranular corrosion (IGC) experiment is conducted in accordance with the provisions of the GB/T 7998-2023 [26] standard. The aged samples are encapsulated with epoxy resin and then polished

with sandpaper to expose the surface. Following alkali and acid washing, the samples are placed in a corrosive medium and soaked at $(30 \pm 3)^\circ\text{C}$ for 6 h before being removed. The corrosive medium is composed of a solution of 57 g/L NaCl and 10 mL/L H_2O_2 .

Electrochemical measurements were performed using a Princeton electrochemical workstation (CH Instruments, Inc., Austin, TX, USA) with a standard three-electrode system. The alloy sample served as the working electrode, with a saturated calomel electrode (SCE) as the reference and a platinum sheet as the counter electrode. All tests were conducted in 3.5 wt.% NaCl solution at room temperature. Samples were polished to a mirror finish prior to testing. Open-circuit potential (OCP) was monitored for 600 s for system stabilization, followed by Tafel polarization tests at a scan rate of 1 mV/s within ± 0.6 V vs. OCP. Each test was repeated in triplicate to ensure data reliability.

The surface corrosion morphology of the soaked sample was captured using scanning electron microscopy (SEM) (FEI CO., Hillsboro, OR, USA), and the cross-section was then polished and observed under a Leica DMILM metallographic microscope (Leica Microsystems Inc., Wetzlar, Germany). Based on these observations, the intergranular corrosion type was determined, and the maximum corrosion depth and the range of corrosion distribution were measured. To ensure the representativeness and reliability of statistical data on intergranular corrosion, each sample must undergo at least 3 independent statistical analyses of corrosion morphology. Furthermore, the selected observation section spacing for each analysis should be at least 3 mm. The thinning samples used for transmission electron microscopy (TEM) (FEI CO., Hillsboro, OR, USA) characterization must be processed on a dual-jet PI-type electrolytic thinning instrument (Ruiling Innovation Technology Co., Ltd., Beijing, China). For the dual-jet electrolytic thinning, the control voltage is 12 V. The electrolyte is a mixed solution of nitric acid and methanol at a volume ratio of 1:3. The electrolyte is maintained at a cooling temperature of -25°C to -35°C .

3. Results and Discussion

3.1. Hardness Behavior

The age-hardening curves of the Al-3.6Cu-1.0Li-0.40Mg-0.32Mn-0.12Zr alloy, examined in the present study under various artificial aging processes, are shown in Figure 2. The applied artificial aging temperature regimes all correspond to T6 heat treatments. The specific aging temperatures used are 150°C , 160°C , and 170°C . Based on the hardness test results, the studied alloy exhibits age hardening under different aging heat-treatment conditions. As aging time increases, the hardness of the three alloys gradually improves. At 2 h of aging, the hardness of Sample-170 is the highest, at about 120 HV. The hardness values of Sample-150 and Sample-160 are roughly the same, at about 108 HV. The higher hardness of Sample-170 at an aging temperature of 170°C for 2 h can reasonably be attributed to the formation of the T1 phase, which provides good strengthening, within the shorter aging time at the higher artificial aging temperature, consistent with the results reported by Jiang [11].

After 8 h of aging, Sample-150 reaches its first hardness peak. However, Sample-160 has reached peak aging, and the hardness gradually increases with aging time. The hardness of Sample-170 reaches its peak value in 24 h. Notably, at lower aging temperatures (150°C), the alloy samples exhibit a second hardness peak after 72 h, following the first. This evolution of hardness with aging temperature should be related to the aging precipitation behavior in the alloy. In addition, the maximum hardness value (132 HV) was obtained for Sample-150 after 72 h of aging. The maximum hardness value (166 HV) was obtained for Sample-160 at 72 h, while the maximum hardness value (168 HV) was obtained for Sample-170 at 72 h. Overall, the hardness values measured for Sample-160 and Sample-170 after a long time of aging were roughly the same, while the hardness value of Sample-150 was the lowest, reduced by about 45 HV compared with the other two sam-

ples. In addition, the age-hardening response rate increases with increasing artificial aging temperature. In this paper, although there are two hardness plateaus at lower aging temperatures, the aging hardness response rate remains the lowest among the maximum hardness values. This does not violate the aforementioned rule governing the evolution of the aging hardening response rate.

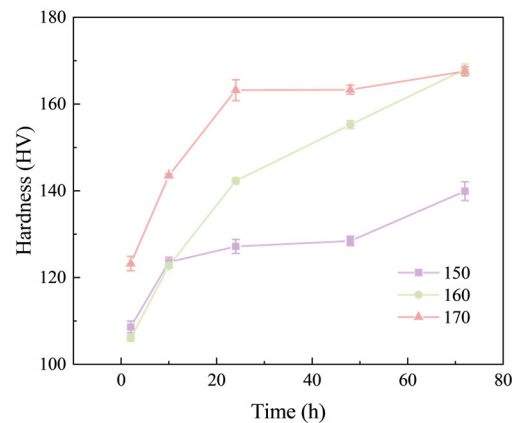


Figure 2. Micro-hardness curves as a function of the aging time for samples aged at 150 °C, 160 °C and 170 °C.

3.2. Intergranular Corrosion Behavior

Unlike the original design purpose of the experimental scheme used in the previous study, the experiment in this paper focuses on the influence of aging temperature on the corrosion mode in the same Al-Cu-Li alloy and on the evolution of corrosion behavior during the full-aging process. In the Al-Cu-Li alloy studied, three corrosion modes can be identified: full intergranular corrosion, localized intergranular corrosion, and pitting [27]. It must be distinguished that when intergranular corrosion and pitting are superimposed, it is necessary to determine whether the pitting is derived from the decrease in grain binding force caused by intergranular corrosion or directly generated by the corrosion of the intermetallic compound. In the present study, the former is referred to as intergranular corrosion with pitting, and the latter is defined as pitting.

During the aging of the studied alloy, as aging time increases, the corrosion behavior evolves from intergranular to pitting, then to local intergranular corrosion. In the current-aged alloy, the main forms of corrosion are intergranular corrosion and pitting corrosion, which can be further refined into comprehensive intergranular corrosion with pitting corrosion, local intergranular corrosion with pitting corrosion, and standard pitting corrosion. Based on the team's previous research, the morphology of intergranular corrosion can be divided into four types: general IGC, local IGC, pitting corrosion, and simultaneous pitting corrosion and intergranular corrosion. In this study, the alloy cross-section exhibits a continuous intergranular corrosion network and a distinct intergranular corrosion morphology, as shown in Figure 3c. Although local intergranular corrosion also exhibits an obvious intergranular morphology, it is limited to local areas. Pitting is characterized by a large cross-sectional area of the opening that gradually decreases from the surface to the interior. Pit corrosion and intergranular corrosion also occur simultaneously. Intergranular corrosion is preferentially distributed at the edge of the corrosion pit, forming an intergranular corrosion grid around it.

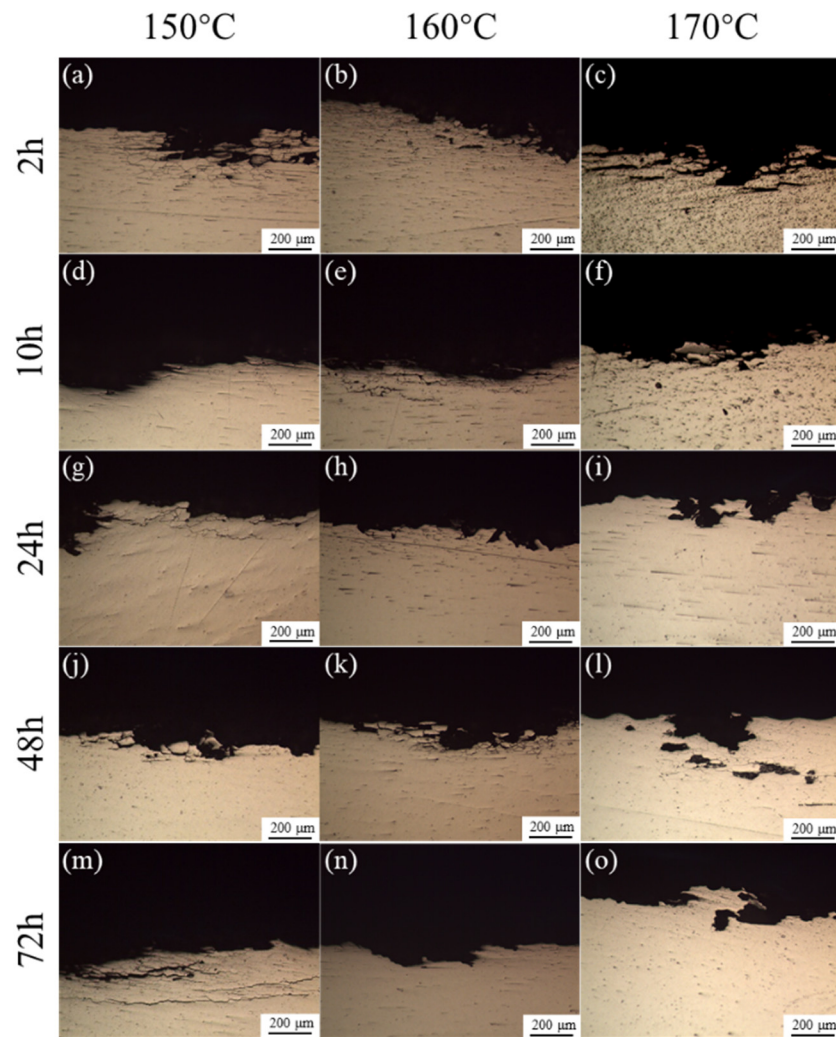


Figure 3. The typical corrosion morphologies of samples aged at 150 °C, 160 °C, and 170 °C for 2 h (a–c), 10 h (d–f), 24 h (g–i), 48 h (j–l), and 72 h (m–o).

The results show that after aging for 2, 10, 24, 48, and 72 h at 150 °C, the Al-Cu-Li alloy exhibits synchronous intergranular and pitting corrosion. By comparing the corrosion morphologies at different aging times, it can be seen that the intergranular corrosion morphologies of Sample-150 are more obvious after aging for 2 h, 10 h, and 72 h; the pitting depth after 72 h of aging is relatively shallow. The corrosion section of Sample-160 shows obvious intergranular corrosion and pitting after aging treatments of 2 h, 10 h, 24 h, and 48 h. The intergranular and pitting corrosion are mixed after 48 h of aging treatment; the corrosion section of Sample-160 after 72 h of aging treatment shows only pitting corrosion morphology. It is worth noting that Sample-160, after aging for 10 h, shows severe intergranular corrosion with a large corrosion depth. Sample-160, after 72 h of aging, has the maximum corrosion depth among all aging times at this temperature. Sample-170 shows obvious intergranular and pitting corrosion with a mixed morphology after aging for 2 h. When the aging time is extended to 10 h, the corrosion pattern shifts from comprehensive intergranular corrosion to local intergranular corrosion and pitting corrosion, indicating that the continuity of intergranular corrosion is weakened. As aging time was further increased to 24 h, 48 h, and 72 h, the corrosion section of Sample-170 exhibited only pitting corrosion morphology. It is worth noting that the corrosion depth of Sample-170 reaches its maximum after 24 h of aging treatment, indicating that although there is no obvious intergranular corrosion at this time, the longitudinal development of pitting corrosion is most severe.

The corrosion type, average and maximum depths of intergranular corrosion, and average and maximum depths of pitting corrosion of the studied Al-Cu-Li alloy after aging at different temperatures are shown in Table 2. In general, the corrosion evolution of the studied Al-Cu-Li alloy can be divided into the following four stages: local intergranular corrosion → comprehensive intergranular corrosion → local intergranular corrosion → pitting corrosion. As the aging temperature increases, the corrosion rate of each alloy clearly accelerates, and the alloy's sensitivity to intergranular corrosion gradually decreases.

Table 2. Dominant corrosion mode, IGC depth, max IGC depth, pitting corrosion depth, and max pitting depth of the studied Al-3.6Cu-1.0Li-0.40Mg-0.32Mn-0.12Zr alloys.

Aging-T6		Dominant Corrosion	IGC Depth	Max IGC Depth	Pitting Corrosion Depth	Max Pitting Depth
Temperature	Time (h)	Mode	(μm)	(μm)	(μm)	(μm)
150 °C	2	Local IGC with pitting	256.02	342.86	143.11	295.21
	10	Local IGC with pitting	184.07	305.22	88.11	203.34
	24	Local IGC with pitting	258.39	417.76	203.59	377.37
	48	Local IGC with pitting	233.04	317.16	134.06	272.72
	72	Local IGC with pitting	207.22	313.25	69.96	144.57
160 °C	2	Local IGC with pitting	327.39	452.92	181.82	385.36
	10	General IGC with pitting	234.12	371.97	134.36	265.51
	24	Local IGC with pitting	237.27	387.96	222.16	368.68
	48	Local IGC with pitting	207.16	256.10	97.56	202.44
	72	Pitting	---	---	56.14	168.92
170 °C	2	General IGC with pitting	449.10	495.13	262.62	412.23
	10	Local IGC with pitting	339.83	481.02	198.19	405.07
	24	Local IGC with pitting	294.99	386.24	149.41	294.11
	48	Pitting	---	---	140.28	274.83
	72	Pitting	---	---	134.84	239.02

In this paper, the intergranular corrosion behavior measured in Figure 3 is statistically analyzed. The characteristic data for intergranular and pitting corrosion at the critical maximum and average corrosion depths are combined with the corrosion morphology evolution law to form curves, facilitating interpretation of the corrosion behavior law (as shown in Figure 4). As shown in Figure 4, the corrosion depth remains shallow with increasing aging time, regardless of aging temperature. However, it can also be observed that the corrosion behavior follows a characteristic law that varies with aging temperature. At an aging temperature of 170 °C and an extended aging time, the corrosion depth decreases significantly, and the corrosion type changes accordingly. Intergranular corrosion occurs only 48 h before aging, which is significantly different from the corrosion behavior at the other two aging temperatures. At the same time, when the aging temperatures are 160 °C and 150 °C, the corrosion depth first decreases, then increases, and then decreases again. In this process, the type of corrosion that is always present is intergranular corrosion with pitting. During prolonged aging, a similar corrosion evolution law is observed. It should be noted that at lower aging temperatures of 150 °C and 160 °C, the evolution of intergranular corrosion depth in the investigated alloy samples coincides with the onset of pitting. At the higher aging temperature of 170 °C, this rule does not appear at the same time point; instead, it is distributed across different aging time points.

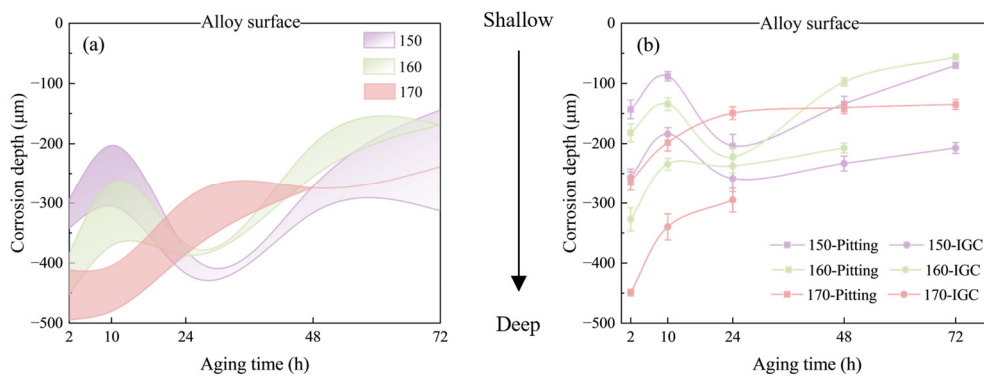


Figure 4. Corrosion depth of the studied Al-3.6Cu-1.0Li-0.40Mg-0.32Mn-0.12Zr alloys aged at different temperatures and times: (a) max depth; (b) average depth.

3.3. Corrosion Surface Evolution Characteristics

To more comprehensively assess the corrosion behavior of the studied samples at different aging temperatures, metallographic images of longitudinal sections across a wide size range (as shown in Figure 5) are combined to clearly illustrate the corrosion characteristics. As shown in Figure 5, different aging temperatures yield distinct macroscopic corrosion morphologies after 24 h of artificial aging. It can be seen that although all three samples exhibit typical intergranular corrosion with pitting, the corrosion features differ at the small scale. As the aging temperature increases, the probability of pitting increases. As the aging temperature decreases, the probability of intergranular corrosion with pitting increases significantly. However, at low aging temperatures, pitting corrosion over a wide range, accompanied by the shedding of flake grains, can lead to a significant increase in corrosion depth.

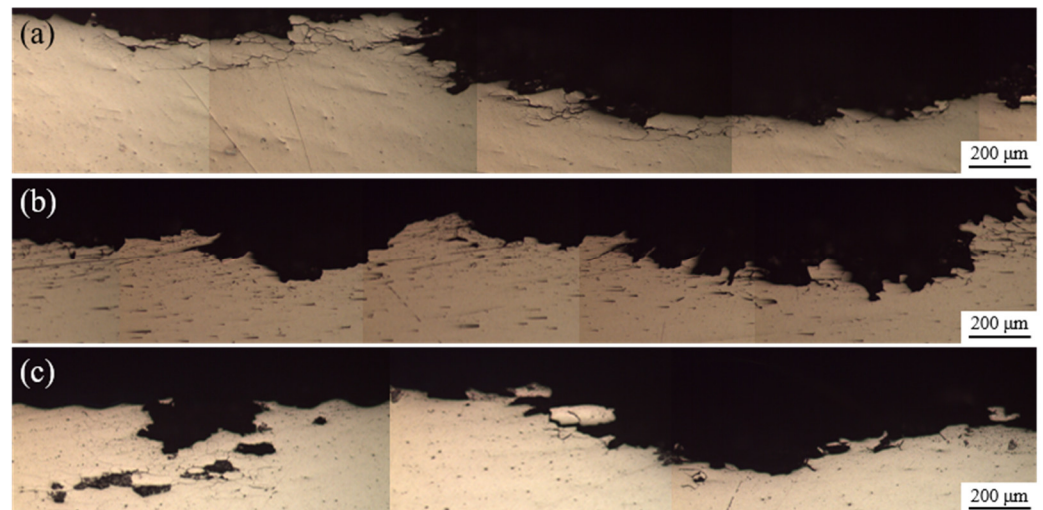


Figure 5. The typical corrosion morphologies of the whole corrosion section of Sample-150 (a), Sample-160 (b), and Sample-170 (c) aged for 24 h.

To further study the surface corrosion initiation and corrosion development characteristics of samples treated with different aging temperatures, it is necessary to conduct a detailed characterization of the alloy's surface corrosion morphology and to study and analyze the correlation between the origin points that may cause corrosion and the system. Figure 6 shows SEM images of the corrosion surface morphology of the studied Al-Cu-Li alloy after 24 h of aging treatment at 150 °C (Figure 6a–a2), 24 h of aging treatment at

160 °C (Figure 6b–b₂), 24 h of aging treatment at 170 °C (Figure 6c–c₂), and 72 h of aging treatment at 170 °C (Figure 6d–d₂). As shown in Figure 6a₂,b₂, the Al-Cu-Li alloy shows the mixed morphology of local intergranular corrosion and pitting corrosion, which is consistent with the morphology in Figure 3g,h. In addition, the grain boundary outline in Figure 6b₂ is more obvious, indicating that the pitting corrosion is more serious under the aging condition of 160 °C, which is consistent with the statistical data in Table 2. In Figure 6c,d, no obvious grain boundary features are observed on the Al-Cu-Li alloy surface; instead, a fierce pitting morphology is observed, which is also consistent with the statistical data in Table 2.

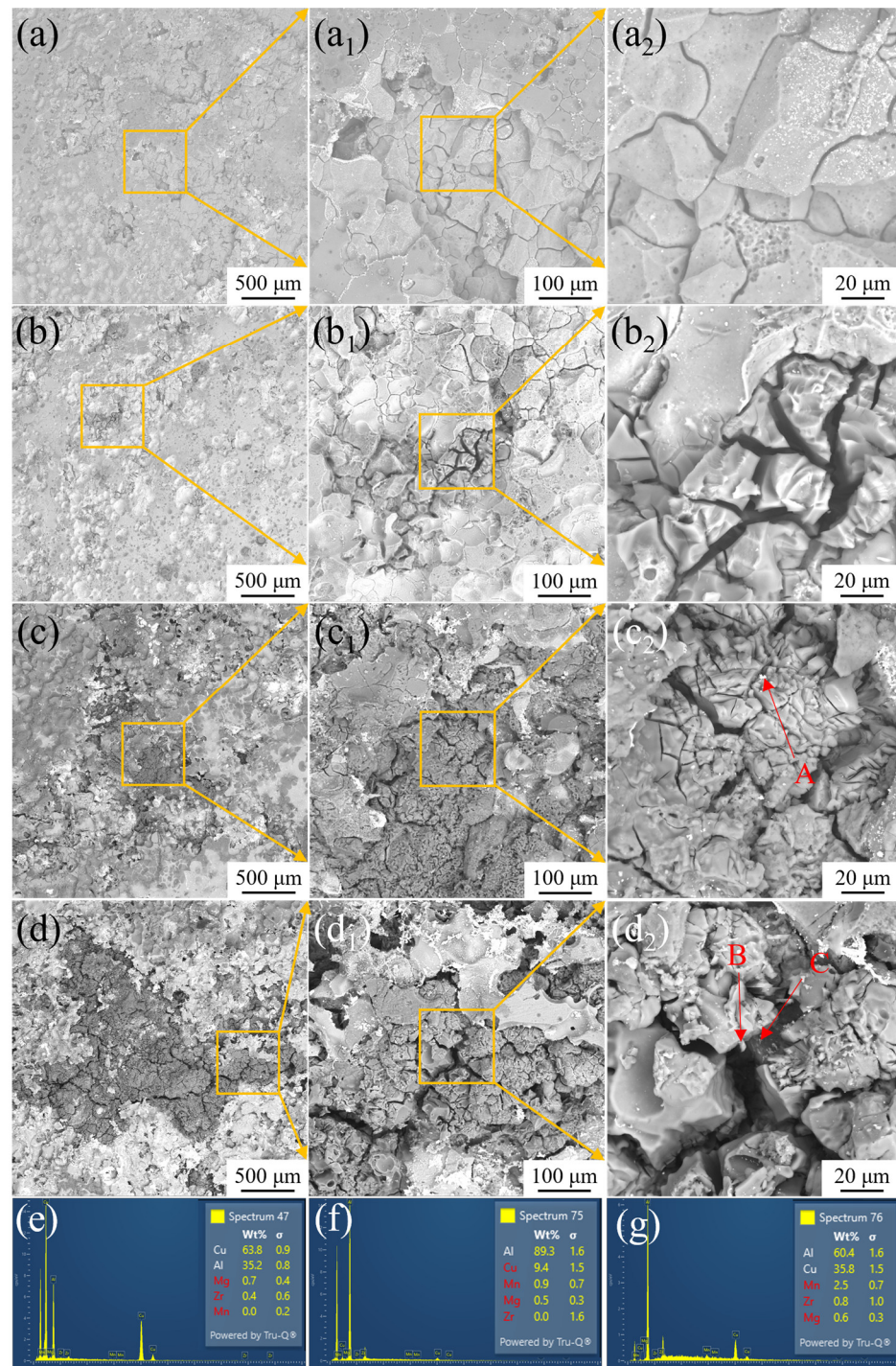


Figure 6. Surface corrosion morphology of Sample-150 aged at 150 °C for 24 h (a–a₂), Sample-160 aged at 160 °C for 24 h (b–b₂), Sample-170 aged at 170 °C for 24 h (c–c₂) and 72 h (d–d₂); EDS of A (e), EDS of B (f), and EDS of C (g).

It is worth reiterating the results from the previous text, which indicate that the hardness values differ significantly after 24 h of aging across the three conditions. In terms of hardness response, the values increase significantly with rising aging temperature. At the same aging time point, although the corrosion types are similar, the corrosion depth gradually increases as the aging temperature decreases. This indicates a concurrent improvement trend between hardness and corrosion resistance at this aging stage, which differs from the commonly reported inverse trade-off between hardness and corrosion properties. From the corrosion morphology and point scan analysis results shown in Figure 6, it can be observed that as the aging temperature increases, the number and probability of corrosion initiation points increase. Specifically, in macroscopic corrosion morphology, the probability of deep gully-like features increases significantly. However, at 24 h of aging, the corrosion depth did not increase correspondingly. Specifically, as the aging temperature increases, the probability of corrosion initiation increases, whereas the driving force for corrosion expansion may not increase.

From the EDS point-scanning analysis of corrosion initiation sites, the distribution of detectable elements (Al, Cu, and Mg) was obtained. Considering that conventional EDS cannot reliably detect the Li element, the phase assignment is based on the well-established precipitation sequence of Al-Cu-Li-Mg alloys and previous reports in the literature [11,13,28], and the intermetallic compounds at corrosion initiation sites are inferred to be Al₂Cu, Al₂CuLi, and Al₂CuMg. At lower aging temperatures, Al₂Cu particles are more likely to serve as corrosion initiation sites. As the aging temperature increases, the probability of Al₂CuLi and Al₂CuMg serving as corrosion initiation points increases. Notably, at an aging temperature of 170 °C, the probability of Al₂CuMg acting as corrosion initiation sites rises with increasing aging time. This effect is also speculated to be the reason for the transition in corrosion mode from intergranular corrosion to pitting corrosion under these conditions.

3.4. Corrosion Electrochemical Behavior

The electrochemical corrosion test provides valid evidence for studying the corrosion behavior of alloy materials [24,29,30]. The experimental results of this test will also provide robust data to predict the corrosion development trend of the studied alloy materials. In this paper, the open-circuit potentials and Tafel curves of the investigated samples at different aging times are summarized. The corrosion potential corresponding to the open-circuit potential in the statistical curve will reflect the alloy's corrosion initiation mechanism. In contrast, the summary and evolution rules of the self-corrosion potential in the Tafel curve will help explain the alloy's corrosion propagation dynamics. By summarizing data on corrosion potential, self-corrosion potential, and corrosion resistance, it is possible to explain the evolution law of the samples studied in this paper at different aging time points.

Figure 7 and Table 3 present the electrochemical results of the studied alloy under different aging temperatures and times, including Tafel polarization curves, corrosion current density and open-circuit potential. As shown in the figure, the corrosion current of the Al-Cu-Li alloy fluctuates after aging for different durations, and the fluctuation period shortens with increasing aging temperature. The figure also shows that the self-corrosion potential of the three alloys studied decreases gradually with increasing aging time. Among them, the self-corrosion potential of Sample-150 increases slightly from aging to over-aging (72 h). The same slight increase in self-corrosion potential during the over-

aging stage also occurs in Sample-170. However, this phenomenon does not appear in Sample-160. With increasing aging time, the self-corrosion potential of the alloy gradually decreases, and its decline accelerates in the over-aging stage. At the same time, the trend of open-circuit potential for the three different artificial-aging heat-treated samples with aging time is consistent with the change rule of self-corrosion potential. The consistency between the two corrosion potentials provides an effective basis for interpreting the law governing corrosion behavior. The corrosion behavior law corresponding to the above electrochemical evolution characteristics of corrosion is that, at low aging temperatures, the corrosion depth first increases and then decreases. When the aging temperature is highest, the corrosion depth decreases gradually, but the corrosion type changes. At an aging temperature of 160 °C, the corrosion depth first decreases, then increases, and then decreases again.

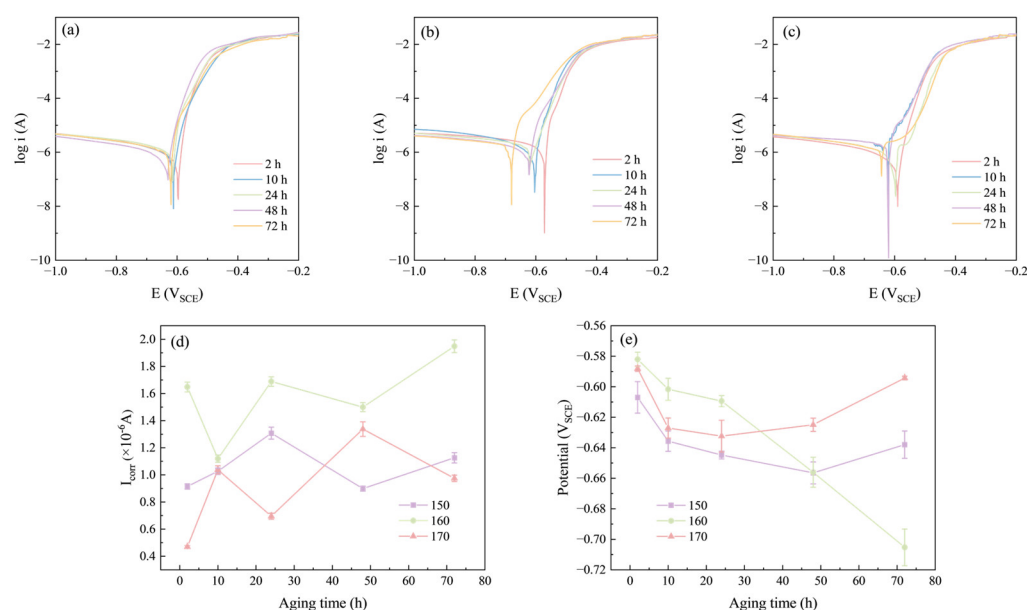


Figure 7. Tafel curves of Al-Cu-Li-Mg alloys aged at 150 °C (a), aged at 160 °C (b), and aged at 170 °C (c); corrosion current (d); and OCP (e).

Table 3. Open-circuit potential (OCP), E_{corr} , I_{corr} , and polarization resistance (R_p) determined from the Tafel curves of the studied alloy in different aging conditions.

Sample	Aging Time (h)	Potential (V_{SCE}) (V)	I_{corr} (A/cm ²)	R_p (Ω ·cm ²)
Sample-150	2	−0.60699	9.14682×10^{-7}	13963.5
	10	−0.63567	1.02698×10^{-6}	9876.2
	24	−0.64477	1.30769×10^{-6}	7118.4
	48	−0.65648	8.98783×10^{-7}	12898.0
	72	−0.63795	1.12602×10^{-6}	4326.0
Sample-160	2	−0.58205	1.6481×10^{-6}	5551.6
	10	−0.60163	1.11937×10^{-6}	10457.3
	24	−0.60939	1.68824×10^{-6}	8013.4
	48	−0.65606	1.49945×10^{-6}	9711.0
	72	−0.70528	1.94882×10^{-6}	10867.3
Sample-170	2	−0.58812	4.69212×10^{-7}	36764.1
	10	−0.62707	1.03751×10^{-6}	6849.2
	24	−0.63238	6.95232×10^{-7}	7310.7
	48	−0.62492	1.33769×10^{-6}	3319.1

72

−0.59426

 9.74848×10^{-7}

6246.2

To explain the above experimental phenomena, it is not entirely scientific to judge solely by changes in corrosion potential [22,23,31]. This requires introducing an analytical path for polarization resistance, which has been widely used in previous studies.

The sample aged at 170 °C for 2 h exhibits an exceptionally high polarization resistance ($R_p = 36764.1 \Omega \cdot \text{cm}^2$). This abnormally high R_p is closely associated with the accelerated early precipitation of the T1 phase at elevated temperatures. A higher aging temperature significantly increases the diffusion rate of Cu and Li solute atoms in the matrix. This promotes the rapid and uniform nucleation of the fine T1 phase at the initial aging stage [13,32]. The uniformly distributed early precipitates narrow the electrochemical potential difference between grain interiors and grain boundaries. They weaken the galvanic driving force for intergranular corrosion initiation and thus substantially improve the overall polarization resistance of the alloy.

Based on the above technical path analysis, it can be concluded that when the aging temperature is low, the polarization resistance first decreases and then increases. As the aging temperature increases, the polarization resistance first decreases and then increases. At an aging temperature of 160 °C, the polarization resistance gradually decreases. These regular characteristics are consistent with corrosion behavior, and it has been reported that the corrosion behavior of the alloy in the current research is mainly affected by the corrosion expansion power. During the corrosion process, the polarization resistance, which hinders the expansion of corrosion, can be used to explain the law governing corrosion behavior. Through an in-depth correlation between the above electrochemical corrosion behavior and the corrosion behavior results, it can be found that different aging precipitation behaviors lead to distinct electrochemical corrosion behaviors when the current alloy samples are aged at different temperatures. When the aging temperature is lower, as aging time increases, the corrosion depth first increases and then decreases, which is related to the corrosion potential first decreasing and then increasing. When the type of grain-boundary phase remains essentially unchanged, a decrease in the intracrystalline corrosion potential indicates an increase in the potential difference between the grain boundary and the intracrystalline region. This results in increased corrosion kinetics and, consequently, increased corrosion depth. As the aging temperature increases, the corrosion behavior changes. The corresponding corrosion potential first decreases and then increases, but the potential in the later aging period is clearly higher than that at the lower aging temperature. Therefore, it can be speculated that an increase in corrosion potential will lead to a rapid rise in potential in the intracrystalline region, which can reach a level similar to that in the grain boundary region. Therefore, corrosion will be preferentially initiated in the vicinity of particles with positive potential, leading to the formation and linking of pitting corrosion and ultimately large-area corrosion.

3.5. Microstructure Evolution Characteristics

In the current study, to further elucidate the mechanisms underlying corrosion behavior, detailed laws governing microstructure evolution were systematically tested and summarized. The intracrystalline and grain-boundary phase distributions in the peak-aging states of different alloy samples are summarized, and a new feature of corrosion-type conversion is observed at higher aging temperatures between the late-aging and peak-aging stages. Therefore, the microstructural characteristics of peak aging and over-aging at higher aging temperatures are detailed to explain the emergence of this corrosion phenomenon. Figure 8 shows the microstructural characteristics of the intracrystalline precipitated phases at different aging temperatures, and speckle diffractograms were used to assist in their qualitative analysis. In addition, Figure 9 shows the grain-boundary phase

distribution of the three samples at different aging temperatures, with face-scanning analysis used to provide detailed identification of the grain-boundary phase types, which were then qualitatively analyzed.

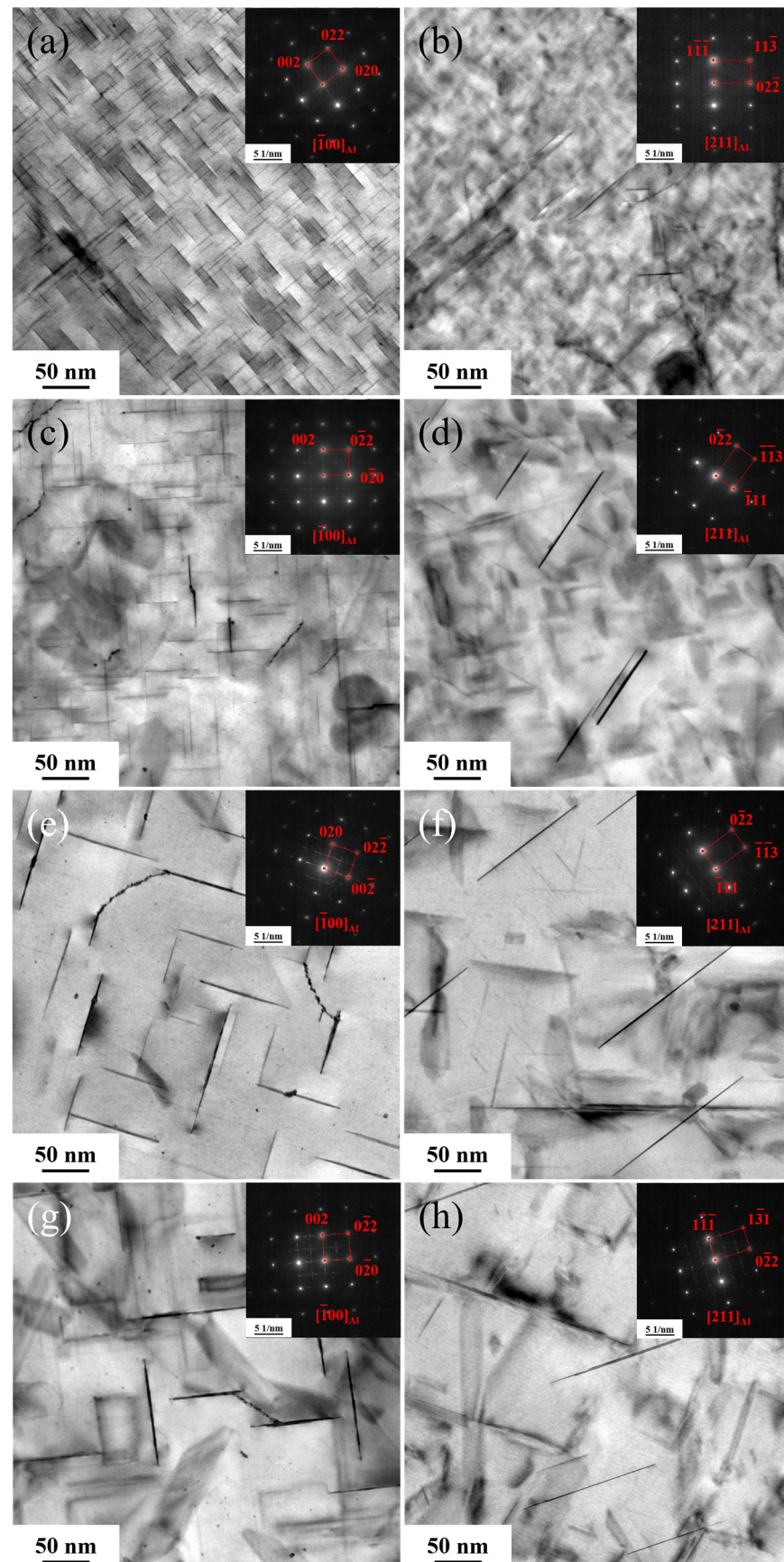


Figure 8. BF-TEM and SAED images of the studied Al-Cu-Li alloys with different aging temperatures: (a) aged at 150 °C for 24 h $\langle 100 \rangle_{\text{Al}}$ and (b) $\langle 211 \rangle_{\text{Al}}$; (c) aged at 160 °C for 24 h $\langle 100 \rangle_{\text{Al}}$ and (d) $\langle 211 \rangle_{\text{Al}}$; (e) aged at 170 °C for 24 h $\langle 100 \rangle_{\text{Al}}$ and (f) $\langle 211 \rangle_{\text{Al}}$; and (g) aged at 170 °C for 72 h $\langle 100 \rangle_{\text{Al}}$ and (h) $\langle 211 \rangle_{\text{Al}}$.

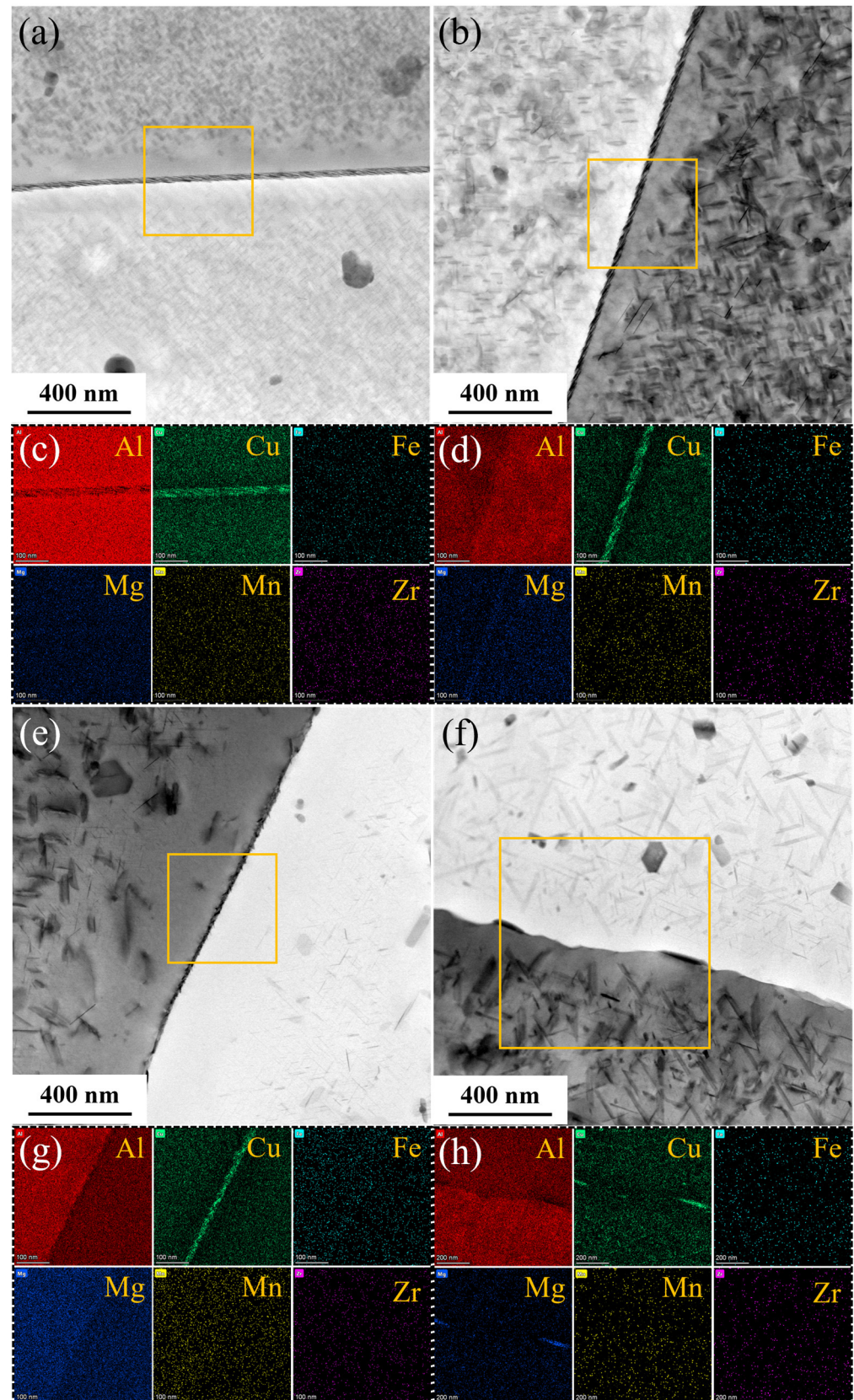


Figure 9. TEM-HADDF images of the studied Al-Cu-Li alloys: (a) Sample-150 aged at 150 °C for 24 h; (b) Sample-160 aged at 160 °C for 24 h; (c) EDS mapping at the GB of Sample-150 aged for 24 h; (d) EDS mapping at the GB of Sample-160 aged for 24 h; (e) Sample-170 aged at 170 °C for 24 h; (f) Sample-170 aged at 170 °C for 72 h; (g) EDS mapping at the GB of Sample-170 aged for 24 h; (h) EDS mapping at the GB of Sample-170 aged for 72 h.

From the SAED diffraction spot diagrams shown in Figure 8, with the analysis from the two dimensions of $\langle 001 \rangle_\alpha$ incidence angle and $\langle 112 \rangle_\alpha$ incidence angle, it can be seen that when the aging temperature is at 150 °C, from the diffraction speckle maps at the $\langle 001 \rangle_\alpha$ incidence angle, the bright streaks in the direction of $\{002\}$ and $\{020\}$ are clearly detected, which is usually associated with the presence of a significant θ'' phase and a certain amount of θ' phase in the studied Sample-150 aged for 24 h. From the diffraction pattern of the $\langle 211 \rangle_\alpha$ incidence angle, it can be seen that there is a weak streak located at the $1/3\{022\}$ direction, which is related to a small amount of T1 phase in the aging sample. When the aging temperature is increased to 160 °C, bright awn lines in the form of dotted lines at intervals in the directions of $\{002\}$ and $\{020\}$ can be seen in the speckle diagram at the $\langle 001 \rangle$ incidence angle, which is related to the presence of a large amount of θ' phase in this aging state sample. At the same time, in the speckled map of $\langle 112 \rangle$ incidence angle, it can be seen that there are obvious awn lines in the $1/3\{022\}$ direction, which is related to the presence of a certain amount of T1 phase in the studied Sample-160 aged for 24 h. When the aging temperature reached 170 °C, the obvious streak located in the direction of $\{002\}$ and $\{020\}$ was found in the view of the $\langle 001 \rangle_\alpha$ incidence angle. At the same time, the T1 phase linked to the slightly streaky spots along the $1/3\{022\}$ direction was identified in Sample-170 aged for 24 h. In addition, the S' phase was determined from the SAED image using the $\langle 112 \rangle_\alpha$ diffraction spot. At the same aging temperature of 170 °C, as the aging time was extended to 72 h, the brightness of the streaks on the diffraction spot diagrams associated with the T1 phase increased, which was related to an increase in the amount of T1 phase in the sample during the same aging state. On the other hand, the brightness of the streaks associated with the S' phase decreases in the diffraction spot diagrams, which is related to the decrease in the number of S' phases present at the aging for more time [28,32,33].

By analyzing selected diffraction spot patterns of the studied alloys aged at different temperatures and combining them with bright-field TEM images, the aging precipitation behavior of the alloys is examined [34]. At low aging temperatures (150 °C), the main precipitated phase is phase θ'' , with a length of about 30–45 nm. A considerable amount of phase θ' is also present, with a size of about 15–30 nm. A small amount of phase T1, about 40 nm, and a very small amount of phase S' are also observed in the aged alloy sample. As the aging temperature increases to 160 °C, the main precipitated phase in the alloy matrix is the θ' phase, with a decrease in the θ'' phase and an increase in the T1 phase. At this aging temperature, the size of phase θ'' decreases to 15–35 nm, the size of phase θ' is about 40–50 nm, and the size of phase T1 increases significantly to about 60–90 nm. When the aging temperature is increased to 170 °C, the main precipitated phases are phases θ' and T1, with a small amount of phase S' . The phase composition and distribution characteristics of both intragranular precipitates and grain boundary phases under all investigated aging conditions are systematically summarized in Table 4.

Table 4. Microstructure evolution of the studied Al-3.6Cu-1.0Li-0.40Mg-0.32Mn-0.12Zr alloys aged at 150 °C, 160 °C, and 170 °C for various times.

Aging-T6	Within Grains			At Grain Boundaries
	Main Phase	Second Phase	Trace Phase	
150 °C-24 h	θ'', θ'	T1	S'	Al ₂ CuLi, Al ₂ Cu

160 °C-24 h	θ'	T1, θ''	Al ₂ CuLi, Al ₂ Cu, Mg enrichment at GB
170 °C-24 h	θ' , T1	S'	Al ₂ CuLi, Al ₂ Cu, Mg enrichment at GB
170 °C-72 h	T1, θ'	S'	Al ₂ CuLi, Al ₂ Cu, Al ₂ CuMg

At the same artificial aging temperature of 170 °C, the number of the main precipitated phase T1 increases with aging time. In contrast, the precipitation ratio of the S' phase decreases. During this aging process, the size and length of the main precipitated phase T1 do not change significantly, remaining between 50 and 80 nm, while the trend of increasing phase width is more pronounced. This indicates that, at higher aging temperatures, the T1 phase remains a stable precipitate. In contrast, the microstructural evolution that transforms the θ'' and S' phases into the T1 phase occurs. This phase transformation mechanism has been reported in previous studies [28,32,34,35] and is consistent with the discussion in this paper. This can be used to explain the underlying mechanism of the corrosion-type transformation described in the current paper.

3.6. The Correlation Between Microstructure and Corrosion Behavior at Different Aging Temperatures

Because grain boundaries are a key pathway for the initiation and propagation of corrosion in alloys, it is necessary to characterize and analyze the microstructural features of grain boundaries and adjacent regions [35–37]. Essentially, corrosion evolution is dominated by the electrochemical potential difference between the grain interior and grain-boundary, which is jointly determined by intragranular precipitate types, grain boundary precipitate characteristics, and PFZ width.

Figure 9 shows microstructural images of grain boundaries and adjacent regions in the studied alloy samples under different aging conditions. It was observed that continuous grain-boundary phases formed along the grain boundaries of the studied alloy samples after 24 h of artificial aging at different temperatures. The width of the grain-boundary phase distribution indicates that lower aging temperatures produce more grain-boundary phases and a wider distribution of precipitated phases. At an aging temperature of 170 °C and an extended aging time, the grain-boundary precipitation phase transition is discontinuous, and the grain-boundary phase matures significantly, with a pronounced increase in phase width. In addition, as the aging temperature increases, the PFZ width near the grain boundary increases significantly. At an aging temperature of 170 °C, the PFZ width of the alloy sample is the largest, and compared to 150 °C, it has increased by about 30–60 nm. This increase in PFZ width will create a wider range of grain boundaries adjacent to the corrosion width, significantly reducing the corrosion current and slowing the corrosion kinetics. This is closely related to the macroscopic phenomenon of the lowest corrosion depth of Sample-170 after 24 h of artificial aging.

Analysis of EDS mapping of grain-boundary phases shows that in Sample-150 with a lower aging temperature, intermetallic compounds containing Cu are present on the grain boundaries. Combined with the characteristics of intragranular precipitation behavior and the typical grain-boundary precipitation features of Al-Cu-Li alloys reported in the literature [11,28], it should be inferred that this alloy phase is a mixed phase of Al₂Cu and Al₂CuLi. Among them, the proportion of the Al₂Cu phase or Cu-containing phase is higher. These noble Cu-rich grain-boundary phases act as cathodes relative to the adjacent anodic PFZ, accelerating continuous intergranular corrosion along grain boundaries. In Sample-160 and Sample-170, Al₂Cu and Al₂CuLi coexist, and there is a certain width of Mg element segregation zone at the grain boundary. Of course, as the aging temperature gradually increases, the number of Al₂CuLi phases on the grain boundary increases, and at the same time, the width of the Mg-containing segregation region near the grain boundary increases. At an aging temperature of 170 °C, the Al₂CuMg phase forms at grain

boundaries as aging time increases. This indicates that as aging time increases, Mg segregation gradually concentrates at grain boundaries and participates in the formation of Mg-containing alloy compounds, consistent with the solute-depletion mechanism of PFZ. The formation of Al_2CuMg phase and the discontinuous distribution of grain boundary precipitates break the continuous cathodic network of grain boundaries, which is the core microstructural origin of the corrosion mode transition.

The microstructural evolution law and corrosion behavior of the alloy samples studied in this article, which were treated at different aging temperatures, are correlated [13,38–40]. This correlation, combined with the laws of microstructural evolution, forms a schematic diagram, as shown in Figure 10. As shown in the figure, in the under-aging stage, the proportion of the T1 phase in the main precipitate phase of the alloy matrix increases significantly with rising aging temperature, which is closely related to the highest corrosion depth of the studied sample at 170 °C. The Al_2Cu and T1 phases coexist at the grain boundary, and as the proportion of the T1 phase increases, this leads to a large potential difference, resulting in a faster corrosion initiation rate.

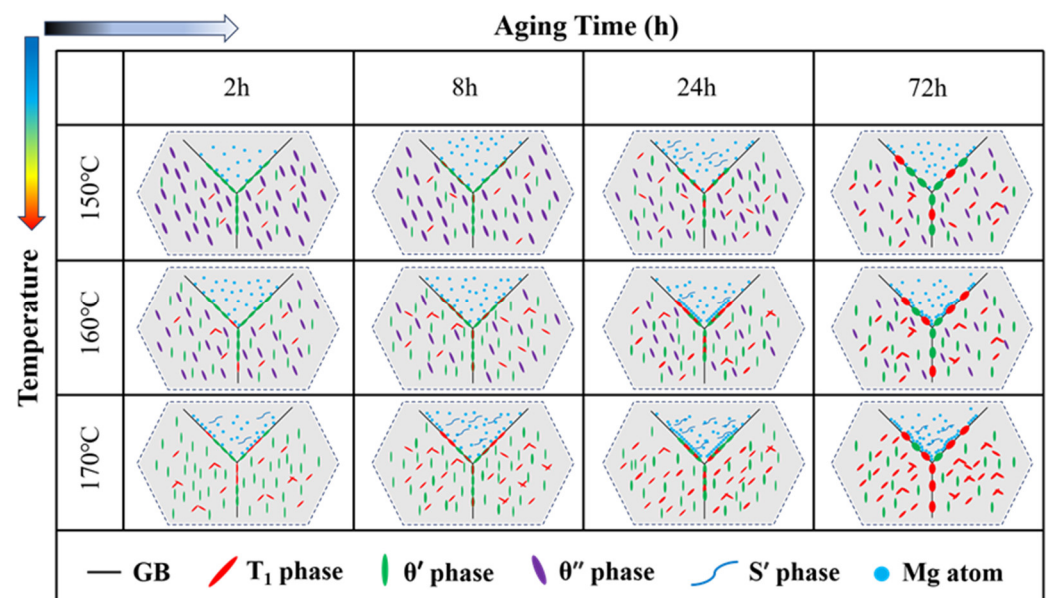


Figure 10. Schematic diagram of the mechanism of the relationship between microstructure and corrosion behavior of the studied alloy.

At 24 h of aging, the corrosion depth increases with decreasing aging temperature, a trend attributed to the presence of the positive-potential phase θ'' in the microstructure. The increase in the proportion of this phase, combined with a narrower PFZ, will lead to preferential corrosion at the grain boundary and in the narrow corrosion channel, resulting in a larger corrosion current and stronger corrosion dynamics [9,20,41]. This explains why the corrosion depth is larger at lower aging temperatures after 24 h of aging.

At a higher aging temperature of 170 °C, the corrosion mode transitions from intergranular corrosion to pitting corrosion with prolonged aging time. This is inferred to be associated with the concentration of Mg near grain boundaries and the decrease in the number of S-phase particles in Sample-170. When the coarse S phase forms at the grain boundary, it forms a relatively negative potential combination with the T1 phase and is preferentially corroded away, thereby forming larger corrosion initiation points. This mechanism accounts for the dominant pitting corrosion morphology at high aging temperatures in the later aging stage.

4. Conclusions

In the current paper, the correlation between the microstructural evolution characteristics and the distinguishing corrosion behavior of the same Al-3.6Cu-1.0Li-0.40Mg-0.32Mn-0.12Zr alloy is established by studying the influence of different time-effect heat treatments on the microstructure and corrosion behavior. The main research conclusions are as follows:

- (1) Within the aging temperature range of 150–170 °C, the main precipitates are θ'' , θ' , and T1. The secondary precipitates are S'. The main precipitates at the grain boundaries are θ' and T1, with a small amount of S'. There is segregation of Mg elements in the vicinity of the grain boundaries. As the aging temperature increases, the main precipitates evolve from θ'' and θ' to θ' and T1. Correspondingly, the intermetallic compounds present at grain boundaries and adjacent regions transform from Mg-containing segregation regions to the S' phase.
- (2) The hardness of the alloy increases gradually with extended aging time at all three temperatures. High hardness values are observed even during short-term aging at higher temperatures; this is primarily attributed to the formation of the T1 phase, which provides significant strengthening within a short aging period. Two hardness peaks appear at the lower aging temperature of 150 °C, which are related to the precipitation of the T1 phase and the increased fraction of the S' phase, respectively. The aging hardening response rate increases with rising artificial aging temperatures.
- (3) As the aging temperature increases, the corrosion depth gradually increases in the under-aging stage. After 24 h of aging, as the aging temperature increases, the corrosion depth gradually decreases. At the high aging temperature of 170 °C, a transition from intergranular corrosion to pitting corrosion occurs. For the alloy aged at 160 °C, the corrosion depth first increases and then decreases with increasing aging time. Within the temperature range of 160–170 °C, the corrosion depth decreases continuously with increasing aging time.
- (4) In the under-aging stage, the maximum corrosion depth increases with rising temperature; this is closely related to the significant increase in the proportion of the T1 phase within the matrix's primary precipitation phase. After 24 h of aging, the corrosion depth increases as the aging temperature decreases, which is related to the presence of the positive-potential phase θ'' in the microstructure. At higher aging temperatures, the corrosion mode transition from intergranular to pitting corrosion is inferred to be related to the concentration of Mg near the grain boundaries and a reduction in the amount of the S phase.

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References

1. Wu, G.; Zhang, X.; Zhang, L.; Wang, Y.; Shi, C.; Li, P.; Ren, G.; Ding, W. An insight into the precipitate evolution and mechanical properties of a novel high-performance cast Al-Li-Cu-Mg-X alloy. *J. Alloys Compd.* **2021**, *875*, 159996.
2. Zhao, X.; Liu, W.; Xiao, D.; Ma, Y.; Huang, L.; Tang, Y. A critical review: Crystal structure, evolution and interaction mechanism with dislocations of nano precipitates in Al-Li alloys. *Mater. Des.* **2022**, *217*, 110629.
3. Xu, D.; Li, S.; Jin, X.; Huang, X.; Li, H.; Wang, F. Comparison on the ignition and combustion characteristics of single Al-Li alloy and Al fuel microparticles in air. *Combust. Flame* **2023**, *258*, 113114.
4. Li, S.; Yue, X.; Li, Q.; Peng, H.; Dong, B.; Liu, T.; Yang, H.; Fan, J.; Shu, S.; Qiu, F.; et al. Development and applications of aluminum alloys for aerospace industry. *J. Mater. Res. Technol.* **2023**, *27*, 944–983.
5. Qi, Y.; Zhang, H.; Zhang, W.; Hu, Z.; Zhu, H. Heat treatment of Al-Cu-Li-Sc-Zr alloy produced by laser powder bed fusion. *Mater. Charact.* **2023**, *195*, 112505.
6. Koshino, Y.; Hirose, S.; Aruga, Y.; Shishido, H.; Matsumoto, K. Effects of Cu or Li Addition and Multi-Step Aging Conditions on the Bake-Hardenability of an Al-Mg-Si Alloy. *Mater. Sci. Forum* **2014**, *794–796*, 1152–1156.
7. Huang, J.Y.; Ming, W.Q.; Wu, C.L. Effect of microstructure on corrosion behaviour of Al-Cu-Li alloys. *Mater. Charact.* **2022**, *194*, 112416.
8. Li, Y.Z.; Zeng, G.J.; Lu, D.D.; Liu, Z.Z.; Deng, S.X.; Ma, P.C.; Chen, Y.L.; Zhang, R.F.; Li, J.F. Effect of Sub-Structure and Precipitation Behavior on Mechanical Properties of Al-xCu-Li Alloys. *Met. Mater. Int.* **2023**, *29*, 3204–3221.
9. Huang, J.L.; Li, J.F.; Liu, D.Y.; Zhang, R.F.; Chen, Y.L.; Ma, P.C.; Gupta, R.K.; Birbilis, N. Correlation of intergranular corrosion behaviour with microstructure in Al-Cu-Li alloy. *Corros. Sci.* **2018**, *139*, 215–226.
10. Li, J.F.; Li, C.X.; Peng, Z.W.; Chen, W.J.; Zheng, Z.Q. Corrosion mechanism associated with T1 and T2 precipitates of Al-Cu-Li alloys in NaCl solution. *J. Alloys Compd.* **2008**, *460*, 688–693.
11. Jiang, B.; Wang, H.; Tian, Y.; Yi, D.; Liu, H.; Hu, Z. Effects of aging time on corrosion behavior of an Al-Cu-Li alloy. *Corros. Sci.* **2020**, *173*, 108759.
12. Shruti, M.S.; Panda, P.K.; Nayak, A.K. Recent advances of high entropy materials as corrosion resistant electrocatalysts for sustainable hydrogen production. *Fuel* **2027**, *428*, 140101.
13. Decreus, B.; Deschamps, A.; De Geuser, F.; Donnadieu, P.; Sigli, C.; Weyland, M. The influence of Cu/Li ratio on precipitation in Al-Cu-Li-x alloys. *Acta Mater.* **2013**, *61*, 2207–2218.
14. Tsivoulas, D.; Prangnell, P.B. The effect of Mn and Zr dispersoid-forming additions on recrystallization resistance in Al-Cu-Li AA2198 sheet. *Acta Mater.* **2014**, *77*, 1–16.
15. Gumbmann, E.; De Geuser, F.; Sigli, C.; Deschamps, A. Influence of Mg, Ag and Zn minor solute additions on the precipitation kinetics and strengthening of an Al-Cu-Li alloy. *Acta Mater.* **2017**, *133*, 172–185.
16. Sun, J.; Wu, G.; Zhang, L.; Liu, W.; Zhang, J.; Shi, C.; Li, W. Role of Cu on the mechanical properties and microstructures evolution of Al-xCu-1Li-0.4Mg-1Zn-0.1Zr alloys. *Mater. Sci. Eng. A* **2020**, *792*, 139833.
17. Xu, J.; Deng, Y.; Chen, J. Enhancing the Corrosion Resistance of Al-Cu-Li Alloys through Regulating Precipitation. *Materials* **2020**, *13*, 2628.
18. Yuan, D.; Chen, K.; Chen, S.; Zhou, L.; Chang, J.; Huang, L.; Yi, Y. Enhancing stress corrosion cracking resistance of low Cu-containing Al-Zn-Mg-Cu alloys by slow quench rate. *Mater. Des.* **2019**, *164*, 107558.
19. Liu, S.; Xu, G.; Li, Y.; Tang, L.; Li, H.; Peng, X. Enhanced strength and corrosion resistance of creep-aged AA2050 Al-Li alloy by pre-rolling deformation. *Mater. Des.* **2023**, *226*, 111621.
20. Guérin, M.; Alexis, J.; Andrieu, E.; Blanc, C.; Odemer, G. Corrosion-fatigue lifetime of Aluminium-Copper-Lithium alloy 2050 in chloride solution. *Mater. Des.* **2015**, *87*, 681–692.
21. Dhondt, M.; Aubert, I.; Saintier, N.; Olive, J.M. Effects of microstructure and local mechanical fields on inter-granular stress corrosion cracking of a friction stir welded aluminum-copper-lithium 2050 nugget. *Corros. Sci.* **2014**, *86*, 123–130.
22. Su, W.; Zhu, H.M. In situ corrosion fatigue life of 2198-T8 Al-Li alloy based on tests and DIC technique. *Mater. Test.* **2022**, *64*, 1383–1396.
23. Li, H.; Wang, Z.-W.; Fu, R.; Huang, Y.-C. Influence of aging treatment on corrosion behavior of extruded 2195 Al-Li alloy. *Trans. Nonferrous Met. Soc. China* **2024**, *34*, 755–768.
24. Wang, X.; Li, G.; He, Q.; Jiang, J.; Li, D.; Shao, W.; Zhen, L. Intergranular corrosion of an Al-Cu-Li alloy: The influence from grain structure. *Corros. Sci.* **2023**, *211*, 110845.
25. Bharti, P.; Sahoo, J.R.; Mishra, S. Implementing Strategic Control of T1 and θ' Precipitates for Achieving Superior Mechanical Performance in Al-Cu-Li Alloys. *Metall. Mater. Trans. A* **2026**, *57*, 1366–1373.

26. GB/T 7998-2023; Method for evaluating the susceptibility to intergranular corrosion of aluminum alloys. State Administration for Market Regulation & Standardization Administration of China: Beijing, China, 2023.
27. Li, J.F.; Birbilis, N.; Liu, D.Y.; Chen, Y.L.; Zhang, X.H.; Cai, C. Intergranular corrosion of Zn-free and Zn-microalloyed Al-xCu-yLi alloys. *Corros. Sci.* **2016**, *105*, 44–57.
28. Gao, Z.; Liu, J.Z.; Chen, J.H.; Duan, S.Y.; Liu, Z.R.; Ming, W.Q.; Wu, C.L. Formation mechanism of precipitate T1 in AlCuLi alloys. *J. Alloys Compd.* **2015**, *624*, 22–26.
29. Wang, X.Y.; Jiang, J.T.; Li, G.A.; Wang, X.M.; Shao, W.Z.; Zhen, L. Particle-stimulated nucleation and recrystallization texture initiated by coarsened Al₂CuLi phase in Al-Cu-Li alloy. *J. Mater. Res. Technol.* **2021**, *10*, 643–650.
30. Rioja, R.J.; Liu, J. The Evolution of Al-Li Base Products for Aerospace and Space Applications. *Metall. Mater. Trans. A* **2012**, *43*, 3325–3337.
31. Xia, L.; Li, Y.; Huang, M.; Liu, J.; Hu, P. Improved corrosion resistance of 2195 Al-Cu-Li alloy through T6I6 interrupted aging. *J. Alloys Compd.* **2023**, *956*, 170055.
32. Zhou, C.; Zhan, L.; Liu, C.; Huang, M. Influence of dislocation morphology on T1 precipitation of an Al-Cu-Li alloy. *J. Alloys Compd.* **2023**, *947*, 169463.
33. Lv, P.; Wang, R.; Peng, C.; Cai, Z. Improved strength and ductility of rapidly solidified 2195 alloy by pre-rolling combined with double aging and interrupted aging. *Mater. Sci. Eng. A* **2023**, *873*, 145023.
34. Dorin, T.; Deschamps, A.; De Geuser, F.; Lefebvre, W.; Sigli, C. Quantitative description of the T1 formation kinetics in an Al-Cu-Li alloy using differential scanning calorimetry, small-angle X-ray scattering and transmission electron microscopy. *Philos. Mag.* **2014**, *94*, 1012–1030.
35. Abd El-Aty, A.; Ha, S.; Hou, Y.; Xu, Y.; Zhang, S.H.; Liang-Liang, X.; Ahmed, M.M. Characterization and constitutive analysis-based crystal plasticity of warm flow and fracture behaviours of 2060 Al-Cu-Li alloy. *J. Mater. Res. Technol.* **2023**, *26*, 1624–1648.
36. Ye, F.; Yu, Y.; Zhang, B.; Rong, J.; He, D.; Han, B.; Ma, X.; Zeng, Y.; Xu, Y.; Wu, S. Influence of pre-stretching at ambient and cryogenic temperatures on dislocation configuration, precipitation behaviour, and mechanical properties of 2195 Al-Cu-Li alloy. *J. Mater. Res. Technol.* **2023**, *22*, 2983–2995.
37. Dai, W.; Jiang, Y.; Yao, J.; Wang, Y.; Cao, F. Simultaneously improving the strength and ductility of an Ag-free 2195 Al-Li alloy by T8 treatment with cryogenic pre-rolling. *J. Alloys Compd.* **2024**, *976*, 173214.
38. Li, G.; Xiao, W.; Li, X.; Wen, K.; Gao, G.; Yan, L.; Li, Y.; Yan, H.; Zhang, Y.; Wang, X.; et al. The Influence of Aging Precipitates on the Mechanical Properties of Al-Li Alloys and Microstructural Analysis. *Metals* **2024**, *14*, 506.
39. Dorin, T.; Donnadiou, P.; Chaix, J.-M.; Lefebvre, W.; De Geuser, F.; Deschamps, A. Size distribution and volume fraction of T1 phase precipitates from TEM images: Direct measurements and related correction. *Micron* **2015**, *78*, 19–27.
40. Zhou, C.; Zhan, L.; Li, H.; Chen, F.; Liu, G.; Yan, D. Mechanism of an acceleration in T1 precipitation kinetics in an Al-Cu-Li alloy by electropulsing. *Vacuum* **2021**, *194*, 110558.
41. Chen, X.; Ma, X.; Zhao, G.; Wang, Y.; Xu, X. Effects of re-solution and re-aging treatment on mechanical property, corrosion resistance and electrochemical behavior of 2196 Al-Cu-Li alloy. *Mater. Des.* **2021**, *204*, 109662.

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