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Effect of Multistage Solid Solution Treatment on Microstructure and Properties of AI-6.5Si-5.5Cu-0.2Zr-0.01Sr-0.06Ti-0.2Y Alloy

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In this study, the impact of multistage solid solution treatment on the microstructure and properties of the Al-6.5Si-5.5Cu-0.2Zr-0.01Sr-0.06Ti-0.2Y alloy was examined using various experimental techniques such as optical microscopy, scanning electron microscopy, x-ray diffraction, transmission electron microscopy, and tensile testing. The findings demonstrated that with an increase in the multistage solid solution temperature and duration, the θ phase (stable Al₂Cu) and intermetallic compounds gradually dissolved into the α -Al matrix, leading to a reduction in the residual phase content. The Si phase's equivalent diameter decreased as the solid solution temperature increased, resulting in the spheroidization of eutectic Si, thereby contributing to dispersion strengthening. Consequently, the gradual rise in dislocation density and sample strength led to an enhancement in tensile strength. The G3 solid solution treatment (470 °C × 2 h + 480 °C × 2 h + 500 °C × 2 h + 510 °C × 2 h + 520 °C × 14 h) demonstrated optimal mechanical properties with the ultimate tensile strength reaching 353.58 MPa and a fracture elongation rate of 9.25%. Moreover, this treatment exhibited superior corrosion resistance, as evident from the intergranular corrosion with a maximum depth of 69.36 μ m, an electrochemical corrosion potential of - 1.1895 V, and a corrosion current density of 1.9965 × 10⁻⁷ A/cm².

Keywords Al-Si-Cu alloy, corrosion resistance, mechanical properties, microstructure, solution treatment

1. Introduction

In the last two decades, the use of casting aluminum alloys has significantly expanded, correlating with an increased demand for improved performance. While numerous new materials have emerged, conventional aluminum alloys remain vital in aerospace and weapon industries due to their stable performance and cost-effectiveness, albeit requiring enhanced performance criteria (Ref 1-4). Al-Si-Cu casting alloys, renowned for their favorable casting properties, low thermalexpansion coefficient, high specific strength, and superior wear resistance, have found extensive applications in the automotive and aerospace sectors (Ref 5). Notably, in the automotive transmission system, the engine's most intricate component, the cylinder head, predominantly employs Al-Si-Cu casting aluminum alloys. However, these alloys exhibit subpar mechanical properties, notably low tensile strength, and limited elongation post-fracture, owing to coarse α-Al dendrites and flake eutectic Si, consequently restricting their wider applications (Ref 6, 7).

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Solid solution treatment stands as a widely employed heat treatment process significantly enhancing alloy properties and microstructure (Ref 8-10). For Al-Si-Cu casting alloys, the multistage solid solution treatment, a primary heat treatment process, closely relates the solubility of alloying elements to the solid solution temperature during strengthening (Ref 11). This treatment produces a supersaturated solid solution, significantly boosting the alloy's strength and hardness (Ref 12). Additionally, it effectively dissolves Al₂Cu phases and other second-phase metal compounds at the grain boundaries, ensuring a more uniform distribution of elements throughout the alloy (Ref 13). The eutectic Si undergoes spheroidization at angular edges, resulting in a uniformly distributed rounded Si phase along the grain boundaries, crucial for diffusion strengthening (Ref 14).

Multistage solid solution treatment facilitates complete dissolution of alloying elements and yields a finer grain structure. Solid solution temperature and holding time significantly determine the mechanical properties of casting aluminum alloy materials and form pivotal components of heat treatment process parameters (Ref 15).Compared with singlestage solid solution, multistage solid solution treatment has stronger ability to control the second-phase, and can control the diffusion rate of solid solution elements in the process, which is a more excellent and accurate method of solid solution treatment. In addition, the multistage solution treatment is also beneficial to improve the roundness of eutectic silicon and ensure good plasticity (Ref 16). Zhou et al. (Ref 17) investigated effect of various multistage solution treatments on the microstructure and properties of cold-extruded Al alloy. On this basis, the multistage solution treatments are further studied to make up for the effect of multistage (larger than 4)

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solution treatments on the properties of the Al alloy. In order to further establish the optimum multistage solid solution treatment on the Al-Si-Cu casting aluminum alloy, the maximum solution temperature was gradually increased and each increase was 10 °C to prevent the maximum solution temperature from being too high. To align with the developmental trajectory of cast aluminum alloys and lightweight automotive components, we have proposed three innovative multistage solid solution regimens, adjusting solid solution temperature and holding time. It is imperative to limit the solid solution treatment temperature to below the polyphase eutectic point to avoid the transitional liquid phase. By reasonably elevating the maximum solution temperature to 520 °C and holding it for 14 h while circumventing the transitional liquid phase, residual phases maximally dissolve into the α-Al matrix. This substantially heightens matrix supersaturation, effectively enhancing overall properties and microstructure of the alloy. Through extensive experimentation, we have identified the optimum multistage solid solution treatment for the Al-6.5Si-5.5Cu alloy, achieving tensile strengths exceeding 350 MPa and post-fracture elongation rates exceeding 9%.

2. Experimental Method

2.1 Material Preparation

The experimental trials employed an independently formulated casting aluminum alloy, denoted as Al-6.5Si-5.5Cu. The alloy's melting and heat treatment procedures were executed as follows: Initially, pure Al (99.97 wt.%) was combined with intermediate alloys comprising Al-Si (20 wt.%), Al-Cu (50.12 wt.%), and Al-Ti-B (5.11 wt.%), all introduced into a graphite crucible within an SG20 resistance melting furnace, set to 900 °C. Once melted, intermediate alloys of Al-Zr (4.11 wt.%), Al-Sr (9.89 wt.%), and Al-Y (10 wt.%) were added. Subsequently, the furnace temperature was reduced to 750 °C after a 3-h hold, followed by the introduction of a Mg block. The molten alloy was then maintained at a constant temperature of 750 °C for 15 min. Following this, the molten material underwent a resting phase and was degassed with C₂Cl₆ for two intervals of 15 min each. Finally, the alloy was poured into preheated molds set at 400 °C. The actual composition of the sample, as depicted in Table 1, revealed a composition of Al-6.5Si-5.5Cu-0.2Zr-0.01Sr-0.06Ti-0.2Y.

2.2 Material Characterization

The samples underwent three distinct solid solution treatments, with parameters detailed in Table 2, followed by water cooling. Subsequently, the samples underwent an 18-h aging treatment at 191 °C. Tensile properties were evaluated at room temperature using the WDW-200G tensile tester, employing a tensile speed set at 1 mm/min. Five tensile specimens were taken from each sample, and the tensile samples are shown in

 Table 1
 Composition content of Al-Si-Cu alloy samples

Elements	Si	Cu	Zr	Sr	Ti	Y	Al
wt.%	6.50	5.50	0.20	0.01	0.06	0.2	Bal.

Fig. 1. The tensile fracture morphology was analyzed through a JEOL-JSM-IT300-type scanning electron microscope (SEM). Additionally, to test the intergranular corrosion resistance, samples were subjected to the GB/T 7998-2005 standard procedure. The corrosion solution comprised 100 ml H₂O + 10 ml H_2O_2 + 57 g NaCl, and samples were immersed for 6 h at 35 °C. Evaluation of the maximum intergranular corrosion depth was conducted using an optical microscope (OM, 4XC-MS). Qualitative analysis of the microstructure in the solid solution state was performed using the SEM. Quantitative analysis of the phases was executed via a Bruker AXS-D8 ADVANCE-type x-ray diffractometer (XRD), employing basic parameters: Cu-K rays with a wavelength of 0.15406 nm, continuous scanning range of 30-120°, scanning speed of 5°/ min, and a data point spacing of 0.02°. After undergoing ion shear thinning, the microstructure of aged samples was observed using a JEM-F200 transmission electron microscope (TEM). Sample preparation involved mechanical grinding to 0.1 mm, followed by electrolytic double spraying and ion thinning.

3. Results and Discussion

3.1 SEM and EDS Analysis

Figure 2 presents distinct morphological traits and Si phase distributions across G1, G2, and G3 samples. Within the figure, the gray areas denote the α -Al matrix, the small black particles represent eutectic Si, and white undissolved phases are evident in all three samples. Detailed observation revealed residual phases predominantly constituted by patchy gray-black bodies and needle-like gray-white structures. EDS qualitative analysis was conducted on these residual phases, specifically on point A with gray-white acicular features and point B with gray-black patches, as shown in Table 3. The primary constituents identified in the residual phases of the gray-white needle-like structures were Al and Cu, pinpointing the insoluble phase at point A as the θ phase (stable Al₂Cu). Contrastingly, the grayblack patchy bodies were a composite of Al₃Zr, Al₃Ti, and Al-Si-Cu-Y four-dimensional compounds. The emergence of this Al-Si-Cu-Y compound was attributed to the large atomic radius and electronegativity of the rare-earth element Y, significantly restricting the degree of solid solution. Wang et al. (Ref 18) proposed an interaction strength model (characterizing the stability and solid solubility of compounds in alloys) between solvent element A and solute element B, which was described by Eq 1:

$$W_{A-B} = \left[(R_A - R_B) / R_B / 0.15 \right]^2 + \left[(N_A - N_B) / 0.4 \right]^2 \quad (\text{Eq 1})$$

where R and N are atomic radius of atomic and electronegativity of element, respectively.

Table 4 presents the interaction values between the Y atom and the investigated alloying elements. Notably, the interactions of Y with Al, Si, and Cu exhibit higher reactivity, with corresponding values of 6.2236, 10.1408, and 8.3344, respectively. These findings indicate a pronounced propensity for Y to interact with these three elements, leading to the formation of a four-dimensional Al-Si-Cu-Y compound.

Furthermore, as depicted in Fig. 2, the eutectic Si phase in the G1 sample exhibits primarily angular edges with a substantial presence of undissolved phases between the grains,

Table 2 Process route of samples solution strengthening

Samples	Specific parameters
G1	470 °C × 2 h + 480 °C × 2 h + 490 °C × 2 h + 500 °C × 18 h
G2	470 °C × 2 h + 480 °C × 2 h + 490 °C × 2 h + 500 °C × 2 h + 510 °C × 16 h
G3	$470 \ ^{\circ}\text{C} \times 2 \ \text{h} + 480 \ ^{\circ}\text{C} \times 2 \ \text{h} + 490 \ ^{\circ}\text{C} \times 2 \ \text{h} + 500 \ ^{\circ}\text{C} \times 2 \ \text{h} + 510 \ ^{\circ}\text{C} \times 2 \ \text{h} + 520 \ ^{\circ}\text{C} \times 14 \ \text{h}$



Fig. 1 Geometry of the tensile test specimens (unit: mm)



Fig. 2 SEM images of aged alloy samples with different solution treatments: (a) G1; (b) G2; (c) G3

Table 3EDS analysis results of residual phases (at.%)

Spots	Al	Si	Cu	Zr	Sr	Ti	Y
A	70.21	1.31	28.15	0.09	0.04	0.18	0.02
B	71.69	11.83	1.69	8.96	0.01	5.06	0.76

leading to segregation aggregation. Conversely, in the G2 sample, a moderate degree of spheroidization is observed at the

edges of the eutectic Si phase, resulting in finer Si phases after dissolution and spheroidization. While the θ phase (stable Al₂Cu) gradually dissolved into the α -Al matrix, there remained a considerable presence of θ phase (stable Al₂Cu) and intermetallic compounds at the sample's grain boundaries due to insufficient solution temperature. In terms of the G3 sample, the eutectic Si phase was predominantly spheroidized into rounded dots, showing a sparse and dispersed distribution along the grain boundaries, with a minor portion migrating into the inner grain. Moreover, the θ phase (stable Al₂Cu) and intermetallic compounds at the grain boundaries were effectively dissolved into the α -Al matrix. Figure 2(a) and (c) depicts the fragmentation and necking of Si particles, showcasing changes in Si particle morphology. This altered morphology contributed to improved sample properties. Despite coarse acicular silicon platelets acting as internal stress raiser in the microstructure and providing potential fracture paths, the modified Si fibers exhibited greater flexibility during solidification. This structural adaptation resulted in somewhat higher ultimate tensile strength and significantly increased ductility values for the alloy (Ref 19, 20).

Figure 3(a)-(c) depicts statistical diagrams representing the equivalent diameter of the Si phase. Observations revealed a gradual decrease in the Si phase's equivalent diameter with increase in solution temperature, indicating a trend of smaller Si

 Table 4
 The value of interaction between Y atom and alloying element

A-B	R _A (A)	R _B (A)	<i>N</i> _A (x)	N _B (x)	W _{A-B}
Al-Y	1.18	1.80	1.61	1.22	6.2236
Si-Y	1.11	1.80	1.98	1.22	10.1408
Cu-Y	1.17	1.80	1.9	1.22	8.3344
Sr-Y	1.91	1.80	0.95	1.22	0.6216
Zr-Y	1.45	1.80	1.33	1.22	1.7560
Ti-Y	1.32	1.80	1.54	1.22	3.8004

phases post-dissolution and spheroidization. Figure 3(d) presents a line graph illustrating the Si equivalent diameter size alongside the area fraction of residual phases. The measured area fractions of residual phases for the three samples were 9.28%, 7.84%, and 6.62%, respectively. This indicated that with the rise in solution temperature, the θ phase (stable Al₂Cu) and intermetallic compounds in the sample progressively dissolved into the α -Al matrix, consequently leading to a reduction in the number of residual phases. Furthermore, the decreasing equivalent diameter and the dispersing distribution of the Si phase also exhibited a correlation with the solution temperature increase.

3.2 XRD Analysis and Dislocation Strength

The XRD patterns and Full Width at Half Maximum (FWHM) of the samples post-solid solution treatment are presented in Fig. 4. The diffraction peak intensity of the θ phase (stable Al₂Cu) decreased with the progressive increase in the solution order, consistent with the findings from SEM microstructure analysis. Figure 4(b)-(d) illustrates the half-peak widths of samples G1, G2, and G3, respectively. The FWHM of the G1 sample was lower compared to that of G2 and G3 samples, indicating a larger sub-crystal size in G1. Although the half-peak width of the G3 sample did not notably differ from that of G2, there was a trend of decreasing values, signifying a subsequent rise in solution temperature and an increase in the sub-crystal size of the G3 sample.



Fig. 3 Equivalent diameter: (a) G1; (b) G2; (c) G3. The variation trend of equivalent diameter and area fraction of Si phase: (d)



Fig. 4 XRD pattern of alloy samples: (a), (b-d) local details of (111), (200), and (220) crystal planes under magnification in (a)

According to XRD analysis spectrum, the lattice strain $(\langle e^2 \rangle^{1/2})$ and dislocation density (ρ) can be calculated by Williamson–Hall model. The half-peak width $(\delta 2\theta)$, the angular position of the main highest diffraction peak (θ_0) , the coherent diffraction region size (d), and the XRD detection ray wavelength (λ) could be described by the following equation (Ref 21, 22):

$$\frac{\left(\delta 2\theta\right)^2}{\tan^2\theta_0} = 25e^2 + \frac{\lambda}{d} \left(\frac{\delta 2\theta}{\tan\theta_0 \sin\theta_0}\right) \tag{Eq 2}$$

The Origin software was used to process the data to obtain a linear relationship, as exhibited in Fig. 5. The *d* and $\langle e^2 \rangle^{1/2}$ values can be attained from the intercept $(25 \langle e^2 \rangle^{1/2})$ and slope (λ/d) of the fitted line, the dislocation density (ρ) can be calculated by this equation (Ref 23):

$$\rho = \frac{2\sqrt{3}\langle e^2 \rangle^{\frac{1}{2}}}{(d \times b)} \tag{Eq 3}$$

where the magnitude of Burgers vector b was 0.286 mm, and the relationship between the dislocation strength (σ_p) could be described according to the following equation (Ref 24):

$$\sigma_{\rho} = M \alpha G b \rho^{1/2} \tag{Eq 4}$$

where M, α and G represent the Taylor factor, numerical factor and shear modulus, and their values were 3.06, 0.24, and 26.9 GPa, respectively (Ref 25-27). The relevant dislocation parameters calculated in Eqs 2, 3 and 4 are listed in Table 5.

Table 5 illustrates that the dislocation density and dislocation strength values of the three samples progressively increase with the elevation of the solution temperature. Notably, the G1 sample exhibited the lowest dislocation density and dislocation strength owing to the limited solid solution of the θ phase (stable Al₂Cu) in the α -Al matrix, resulting in a weaker pinning effect on dislocation movement. Additionally, the spheroidization degree of eutectic Si and the effect of dispersion strengthening at grain boundaries remained insufficient. Conversely, the G3 sample demonstrated the highest dislocation density and dislocation strength, increasing by 9.8% and 4.8%, respectively, compared to the G1 sample. This increase might be attributed to the comprehensive dissolution of θ phase (stable Al₂Cu) into the α -Al matrix, enhancing dislocation resistance. Furthermore, the complete spheroidization of eutectic Si at the grain boundaries significantly contributed to dispersion strengthening. Consequently, the rise in solid solution temperature gradually augmented the dislocation strength of the sample due to the progressive dissolution of θ phase (stable Al₂Cu) and intermetallic compounds into the α-Al



Fig. 5 Fitting relationship between $(\delta 2\theta)^2 / \tan^2 \theta_0$ and $\delta 2\theta / (\tan \theta_0 \sin \theta_0)$: (a) G1; (b) G2; (c) G3

Table 5 Dislocation strength contribution and its parameters

Samples	d, nm	$<\!e^2>^{1/2}$	ρ , (10 ¹⁴ m ⁻²)	σ _ρ , MPa
G1	39.10	7.60E-04	2.35	83.78
G2	36.08	7.19E-04	2.41	84.82
G3	40.12	8.56E-04	2.58	87.81

matrix, accompanied by the gradual specialization and dispersion of eutectic Si, aligning with SEM analysis findings.

3.3 TEM Analysis

To delve deeper into the influence of Y on the aging behavior of Al-Si-Cu system alloys, the G3 sample exhibiting optimal properties was specifically chosen for investigation. Figure 6 presents the TEM image of the G3 sample, revealing various precipitated phases. Conventionally, the precipitation sequence in Al-Si-Cu alloys follows GP (Guinier–Preston) zone $\rightarrow \theta'' \rightarrow \theta' \rightarrow \theta$ (Ref 28), where the GP zone represents the Cu Guinier–Preston phase, while θ'' and θ' denote substable phases, and θ indicates the equilibrium phase.

Figure 6(a) depicts the STEM morphology of the typical precipitated phase along the $[110]_{A1}$ direction in the G3 sample, illustrating the plate-like morphology of the precipitated phase θ' (substable Al₂Cu) with an approximate size of 200 nm, uniformly distributed within grains. Such a distribution pattern facilitated maximum solution strengthening (Ref 29). Fig-

ure 6(b) showcases the HRTEM morphology of the GP zone and θ' phase (substable Al₂Cu) along the [110]_{A1} direction. These θ' phases (substable Al₂Cu) exhibited a coherent orientation relationship with the α -Al matrix, aligning along the interface of $(110)_{\theta r} || \{110\}_{\alpha-Al}$ and in parallel to its wide plane.

Figure 6(c) displays the STEM morphology of other intragranular precipitated phases within the G3 sample, revealing a SADP diffraction pattern characteristic of the precipitated phase, identified as Al_8Cu_4Y (Ref 30). This rare-earth phase assumed a tetragonal crystal system with a size of approximately 2 μ m, representing brittle phases within the material, potentially impacting the alloy's plasticity negatively.

Furthermore, Fig. 6(d) illustrates the STEM morphology at the grain boundaries of the G3 sample, depicting no detectable precipitated phase generation. Consequently, this configuration enhances the corrosion resistance of the alloy. Figure 6(e) presents the STEM morphology at the interface between the Si phase and α -Al within the G3 sample, demonstrating a relatively smooth interface with no evident precipitated phase generation. Additionally, Fig. 6(f) delineates the HRTEM morphology of the Si phase along the [110]_{A1} direction in the G3 sample. The orientation relationship between the Si phase and α -Al matrix lacked coherence with the interface of $(110)_{\theta r}$ ||{110}_{$\alpha - Al$}, effectively enabling dislocation pinning. Post fast Fourier transform (FFT) transformation, the diffraction spots of the Si phase were visible.

Figure 7 primarily displays the Si phase area in the G3 sample for surface scan analysis. The STEM image reveals the presence of a stacking fault phenomenon on the Si phase surface. This occurrence was attributed to the Trace Precipi-



Fig. 6 TEM image of G3 sample: (a) STEM; (b) HRTEM of Precipitates; (c) Al_6Cu_6Y phase; (d) Grain boundary morphology; (e) Si phase; (f) Si phase with the α -Al matrix interface (FFT)



Fig. 7 EDS analysis of precipitate of G3 sample

tation and Refinement Effect (TPRE) mechanism activated by Sr and rare-earth (RE) elements within the sample. Consequently, this mechanism hindered the growth of the Si phase, effectively achieving the modification effect. The surface scan analysis results indicated that the precipitated phase identified in the figure is predominantly characterized as Al_3M (M = Zr, Ti, Y) at the nanometer scale. These nano-level Al_3M entities acted as dislocation inhibitors, contributing to precipitation strengthening, thereby enhancing the overall properties of the sample.

3.4 Electrical Conductivity

Figure 8 depicts the measured electrical conductivity of the sample. The observed changes in conductivity could be attributed to the interaction between mobile electrons and localized strain fields generated by solute atoms, impurities, grain boundaries, and dislocations, which acted as scattering centers for the electrons (Ref 31). Guyot et al. (Ref 32) associated the increase in electrical conductivity with the coarsening and growth of precipitates. Similarly, findings by Liu et al. (Ref 33) suggested that the enhancement in electrical



Fig. 8 Electrical conductivity of the samples

conductivity is facilitated by an increase in the volume fraction of precipitates. It was widely acknowledged that the introduction of lattice distortions due to the dissolution of foreign atoms, crystal dislocations, and thermal vibrations results in elevated electrical resistivity and decreased conductivity (Ref 34).

Within the framework of Matthiessen's rule (Ref 35), the total resistivity can be segmented into several components:

$$\rho_t = \rho_0 + \rho_{gb} + \rho_s + \rho_P + \rho_d + \rho_V \tag{Eq 5}$$

where ρ_0 is the lattice resistivity of the Al matrix; ρ_{gb} is the grain boundary resistivity (Ref 36); ρ_s is the resistivity caused by solute atoms dissolving into the matrix; ρ_P is the resistivity due to precipitated phase; ρ_d is the resistivity caused by dislocation; and ρ_V is the resistivity due to vacancy (Ref 37).

Given the same alloy composition and grain structure, the lattice resistivity of the Al matrix remained constant, while other types of resistivity were influenced by solid solution. E.V. Bobrun's study (Ref 38) highlighted that refining the grain to an ultra-fine scale can enhance the alloy's strength without compromising electrical conductivity, suggesting the negligible impact of grain boundary resistivity. As the three samples underwent identical aging treatment, the resistivity attributed to precipitates remained essentially uniform. Relative to ρ_V , resistivity from ρ_s and ρ_d more effectively influences electron scattering (Ref 39).

Following Matthiessen's rule analysis, ρ_s gradually diminished due to increased solute atom content within the matrix, while ρ_d decreased gradually owing to heightened dislocation density and strength, thereby increasing resistance to electron transport between grains. Consequently, total resistivity gradually decreased. Figure 8 indicates that the conductivity of the G1 sample is the highest at 39.8%, while that of the G2 and G3 samples decreases to 38.1% IACS and 36.4% IACS, respectively. It was evident that the conductivity of the samples progressively declines with rising solid solution temperature, corroborating the findings derived from Matthiessen's rule analysis.

3.5 Tensile Properties and Fracture Morphology

The tensile properties of the samples are assessed and depicted in Fig. 9 by the stress-strain curves. Table 6 presents the tensile strength and elongation after fracture, while Fig. 10 illustrates the morphological characteristics of the tensile fracture. The findings in Table 6 demonstrated a gradual increase in the tensile strength of the samples with increase in solid solution temperature. Notably, the G3 sample exhibited



Fig. 9 Tensile properties of alloy sample

Table 6Tensile property data of aged alloy samples withdifferent solution treatments

Samples	UTS, MPa	Elongation, %	
G1	282.05 ± 3.5	7.20 ± 0.2	
G2	335.71 ± 3.0	8.62 ± 0.3	
G3	353.58 ± 3.6	9.25 ± 0.2	

optimal ultimate tensile strength (UTS) at 353.58 MPa. This enhancement can be primarily attributed to the solubility of the θ phase (stable Al₂Cu). In the G3 sample, the complete dissolution of θ phase (stable Al₂Cu) into the α -Al matrix and subsequent precipitation of minute Al₂Cu during artificial aging facilitated significant precipitation strengthening, contributing substantially to the improved tensile strength of the sample.

Moreover, the considerable rise in tensile strength across samples was also can be attributed to alterations in the size and configuration of eutectic Si particles during the solid solution process. Alongside precipitation hardening offered by Al and Cu, this treatment resulted in the precipitation of sub-microscopic and substable phases containing Si, impeding dislocation motion and thus contributing to precipitation hardening (Ref 40).

Following the attainment of yield strength, the stress in the alloy continued to increase as plastic deformation advances, leading to strain hardening. Two primary hardening mechanisms operate in this process: Δ_{siso} (isotropic strain hardening), caused by changes in dislocation density within the matrix, and Δ_{skin} (kinematic strain hardening), induced by dislocation accumulation in the precipitation phases, grain boundaries, and surrounding the eutectic silica. These mechanisms collectively impeded further strain hardening (Ref 41).

The elongation after fracture of the sample exhibited a gradual increase with the rise in solid solution temperature, where the G3 sample demonstrated optimal elongation after fracture, reaching 9.25%. This enhancement was primarily attributed to the morphological variations in the Si phase. At solution temperatures below 520 °C, the profile of the rareearth-modified fibrous eutectic Si displayed only slight spheroidization due to insufficient solid solution temperatures. However, at 520 °C, the eutectic Si underwent substantial



Fig. 10 Fracture morphologies of aged alloy samples with different solution treatments: (a, b) G1; (c, d) G2; (e, f) G3

spheroidization, aligning with SEM analysis. Moreover, the alteration of eutectic Si particles' morphology, transitioning from the typical needle-like brittle structure to a more rounded form, contributed to improved ductility. Consequently, as the solid solution temperature increased, the sample's UTS and elongation after fracture reached 353.58 MPa and 9.25%, respectively. These enhancements in mechanical properties were a result of the amplified solid solution strengthening effect and the refinement in the morphology and distribution of the residual phase.

Figure 10 depicts the tensile fracture morphology of the sample observed at room temperature. Figure 10(a) and (b)reveals the fracture morphology of the G1 sample. The presence of thick ligamentous nests, a few tearing prongs, and deconstructed surfaces mixed with small areas was evident. These characteristics were indicative of a brittle fracture, with fractures passing through the crystal structure. Figure 10(c) and (d) displays the fracture morphology of the G2 sample. Equiaxed and densely distributed dimples were observed, accompanied by longer tear edges and cleavage surfaces between dimples. The fracture form also indicated a brittle fracture. Figure 10(e) and (f) illustrates the fracture morphology of the G3 sample. The fracture surface displayed numerous small flocculent dimples, small, dense, and rich tearing edges, and relatively fewer cleavage surfaces. This fracture form was characteristic of ductile fracture, indicating that the G3 sample exhibited the highest plasticity among the samples.

3.6 Corrosion Performance Test

Figure 11 illustrates the intergranular corrosion morphology of the samples observed under an optical microscope after standard testing and the polarization curve of the samples in 3.5 wt.% NaCl solution. Table 7 presents the maximum corrosion depth and intergranular corrosion grade of each sample. Upon reviewing the maximum measured depth of intergranular corrosion and corrosion grades, it was observed that the corrosion depth of the G1 sample measures 104.08 μ m, categorized within grade 4. In contrast, both G2 and G3 samples filled within grade 3, with corrosion depths of 99.24 μ m for G2 and 69.36 μ m for G3. The enhanced intergranular corrosion resistance of the G3 sample could be attributed to the reduced presence of residual phases compared to G1 and G2 samples. Consequently, the lower density of microcells in the G3 sample led to improved corrosion resistance (Ref 42). The gradual dissolution of residual phases into the α -A1 matrix in G3 contributed to the enhanced dissolution effect, resulting in its superior intergranular corrosion resistance.

Table 8 presents the corrosion potential ($E_{\rm corr}$) and corrosion current density ($I_{\rm corr}$) results. Among the three samples, the G3 sample exhibited the smallest values for $I_{\rm corr}$ and $E_{\rm corr}$. The $I_{\rm corr}$ was directly proportional to the corrosion rate, implying that higher $I_{\rm corr}$ values indicate poorer corrosion resistance of the sample. Meanwhile, the $E_{\rm corr}$ value signified the tendency and susceptibility of the alloy to corrosion. Analyzing the polarization curve, the G3 sample exhibited the best corrosion resistance at a solution temperature of 520 °C when compared to temperatures of 500 °C and 510 °C. This finding aligned with the experimental outcomes of intergranular corrosion.

4. Conclusions

This study investigated the effects of three multistage solid solution treatments on the Al-Si-Cu casting aluminum alloy (Al-6.5Si-5.5Cu-0.2Zr-0.01 Sr-0.06Ti-0.2Y). The samples which underwent the G3 solid solution treatment (470 °C × 2 h + 480 °C × 2 h + 490 °C × 2 h + 500 °C × 2 h + 510 °C × 2 h + 520 °C × 14 h) demonstrated optimal mechanical properties. The following conclusions were drawn:

(1) With the gradual increase in solid solution temperature, the θ phase (stable Al₂Cu) and intermetallic compounds gradually dissolved into the α -Al matrix, resulting in a reduction in the quantity of residual phases. The



Fig. 11 Corrosion morphology and polarization curve of aged alloy samples with different solution treatments: (a) G1; (b) G2; (c) G3

Table 7	Intergranular	corrosion	grade	of	aged	alloy
samples	with different so	lution treat	ments			

Samples	Maximum corrosion depth, μ m	Rank
G1	104.08	4
G2	99.24	3
G3	69.36	3

Table 8Electrochemical corrosion determination ofsamples in 3.5 wt.% NaCl solution

Samples	$I_{\rm corp}$ A/cm ²	$E_{\rm corr},{ m V}$
G1 G2	$\begin{array}{c} 1.1359 \pm 0.024 \times 10^{-6} \\ 8.4611 \pm 0.094 \times 10^{-7} \end{array}$	-1.1457 ± 0.021 -1.1723 ± 0.035
G3	$1.9965 \pm 0.056 \times 10^{-7}$	$-$ 1.1895 \pm 0.025

microstructure of the G3 sample alloy exhibited the most significant improvement.

(2) As the solid solution temperature gradually rises, the average size of the Si phase also decreased. The spheroidization of eutectic Si gradually exerted a dispersion strengthening effect, while the dislocation density and strength of the samples also increased gradually.

(3) Following the improvement of residual phases and Si phase through the G3 solid solution regimen, the sample exhibited optimal mechanical and corrosion properties. The UTS and elongation after fracture reach 353.58 MPa and 9.25%, respectively. The maximum depth of intergranular corrosion is 69.36 μ m. Furthermore, the electrochemical corrosion potential is measured at -1.1895 V, with a corrosion current density of 1.9965×10^{-7} A/cm².

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Data availability

The data used to support the findings of this study are available from the corresponding author upon request.

Conflict of interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

Reference

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