Microstructure and Low-Cycle Fatigue Behavior of Al-9.0%Si-4.0%Cu-0.4%Mg(-0.3%Sc) Alloy Postprint

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Abstract

To determine the influence of the rare earth element Sc on the low-cycle fatigue behavior of T6 tempered cast Al-9.0%Si-4.0%Cu-0.4%Mg alloy, a comparative study was conducted on the low-cycle fatigue behavior of T6 tempered cast Al-9.0%Si-4.0%Cu-0.4%Mg alloy and Al-9.0%Si-4.0%Cu-4.0%Mg-0.3%Sc alloy. The results indicate that at low applied total strain amplitudes, the Al-9.0%Si-4.0%Cu-0.4%Mg alloy exhibits cyclic strain hardening throughout the entire fatigue deformation period, whereas the Al-9.0%Si-4.0%Cu-0.4%Mg-0.3%Sc alloy displays cyclic strain hardening in the initial stage of fatigue deformation and cyclic stability in the later stage; conversely, at high applied total strain amplitudes, both Al-9.0%Si-4.0%Cu-0.4%Mg(-0.3%Sc) alloys exhibit cyclic strain hardening. The addition of the rare earth element Sc can effectively enhance the cyclic deformation resistance and low-cycle fatigue life of T6 tempered Al-9.0%Si-4.0%Cu-0.4%Mg alloy. At lower applied total strain amplitudes, the cyclic deformation mechanism of T6 tempered Al-9.0%Si-4.0%Cu-0.4%Mg(-0.3%Sc) alloy is planar slip, while at higher applied total strain amplitudes it is a wavy slip mechanism.

Full Text

Microstructures and Low-Cycle Fatigue Behavior of Al-9.0%Si-4.0%Cu-0.4%Mg(-0.3%Sc) Alloy

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Abstract

Al-Si-Cu-Mg cast aluminum alloys possess high mechanical properties and good casting performance. Due to their excellent comprehensive properties, these alloys have found widespread application and have become one of the most important structural materials in the equipment manufacturing industry. In practical engineering applications, many key components are often subjected to alternating loads, making fatigue failure a critical factor concerning the safety and economy of structures across various engineering fields. Although some research on the fatigue behavior of aluminum alloys has been performed, most has focused on understanding general regularities. In particular, the influences of rare earth elements and heat treatment conditions on the low-cycle fatigue behavior of aluminum alloys have not been comprehensively revealed. Investigation of the microstructure and fatigue properties of Al-Si-Cu-Mg cast aluminum alloys can provide not only a theoretical basis for developing new cast aluminum alloys but also a reliable foundation for the safe design and rational use of these materials.

To determine the influence of the rare earth element Sc on the low-cycle fatigue behavior of cast Al-9.0%Si-4.0%Cu-0.4%Mg alloy in the T6 condition, the cyclic stress response behavior, fatigue life characteristics, and cyclic deformation mechanisms of Al-9.0%Si-4.0%Cu-0.4%Mg(-0.3%Sc) cast aluminum alloys under low-cycle fatigue loading were investigated. The results show that at low total strain amplitudes, the Al-9.0%Si-4.0%Cu-0.4%Mg alloy exhibits cyclic strain hardening throughout the entire fatigue deformation process, whereas the Al-9.0%Si-4.0%Cu-0.4%Mg-0.3%Sc alloy shows cyclic strain hardening in the initial stage followed by stable cyclic stress response in the later stage. At higher total strain amplitudes, both alloys exhibit cyclic strain hardening. The addition of Sc can effectively enhance the cyclic deformation resistance and prolong the fatigue life of the T6-treated Al-9.0%Si-4.0%Cu-0.4%Mg alloy. At lower total strain amplitudes, the cyclic deformation mechanism of the T6-treated Al-9.0%Si-4.0%Cu-0.4%Mg(-0.3%Sc) alloys is planar slip, while at higher total strain amplitudes it becomes wavy slip.

Keywords: Al-Si-Cu-Mg alloy, Sc, T6 treatment, low-cycle fatigue, fatigue life, cyclic stress response, cyclic deformation mechanism

1. Introduction

Al-Si-Cu-Mg cast aluminum alloys offer good casting performance, corrosion resistance, and high mechanical properties, and have been widely used in various fields of production and daily life. In recent years, with the rapid development of the aviation and automotive industries, the demand for mechanical properties, especially fatigue performance of cast components, has increased significantly. To ensure safe service of components, a thorough understanding of material fatigue characteristics is required, which has attracted extensive attention from researchers worldwide [1–4].

Cáceres et al. [5] investigated the influence of microstructure on the fatigue behavior of Al-Si series cast aluminum alloys, finding that the secondary dendrite arm spacing (SDAS) of eutectic Si phase significantly affects fatigue life. When SDAS is less than 60 m, fatigue life gradually decreases with increasing SDAS; when SDAS exceeds 60 m, fatigue life slightly increases with SDAS; and fatigue life reaches its minimum at SDAS = 60 m. Wang et al. [6] and Chen et al. [7] found that certain Fe-rich brittle phases (such as Al FeSi, Al Mg FeSi , and Al (Mn,Fe) Si) in A357 cast aluminum alloys significantly reduce fatigue life. During cyclic loading, fatigue cracks tend to initiate at the interface between these brittle phases and the -Al matrix and propagate perpendicular to the applied stress direction along the second phase-matrix interface. The crack propagation rate at the brittle phase-matrix interface is 1.5 times faster than that in uniform microstructures. In contrast, in fine and uniform microstructures, fatigue cracks initiate at the intersection of slip bands with the specimen surface and propagate along slip band directions. Bray et al. [8] summarized the effects of precipitates on fatigue crack propagation behavior in Al-Cu-Mg alloys, identifying a critical diameter d_c for precipitates such as S (Al CuMg) and T (Al CuLi). During fatigue deformation, when precipitate diameters are smaller than d_c, dislocations can shear these precipitates, leading to non-uniform strain distribution and slip localization. During fatigue crack propagation, this slip localization can enhance crack tip shielding effects such as roughness-induced crack closure, crack deflection, and branching, thereby reducing crack growth rates.

This work investigates the strain-controlled low-cycle fatigue behavior of cast Al-9.0%Si-4.0%Cu-0.4%Mg(-0.3%Sc) alloys (compositions in mass fraction, hereafter) to explore the influence of the rare earth element Sc on the low-cycle fatigue deformation behavior of Al-9.0%Si-4.0%Cu-0.4%Mg alloy, providing a theoretical basis for improving the performance of existing alloys and developing new fatigue-resistant cast aluminum alloys.

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2. Experimental Procedures

The materials used were cast Al-9.0%Si-4.0%Cu-0.4%Mg alloy and Al-9.0%Si-4.0%Cu-0.4%Mg-0.3%Sc alloy. The rare earth element Sc was added as an

Al-2%Sc master alloy. Alloys were melted in an SG-5-10 crucible resistance furnace at 740 °C and cast with a mold preheating temperature of 240 °C and pouring temperature of 720 °C.

Cast fatigue specimens were solution-treated and aged (T6) using an SX-4-10 box resistance furnace. The heat treatment process consisted of solution treatment at 520 °C for 8 h followed by water quenching, and aging at 175 °C for 6 h followed by air cooling.

Microstructures were observed using a Neophot-21 optical microscope. The etchant used was 2 mL HF + 3 mL HCl + 5 mL HNO + 250 mL H O.

Low-cycle fatigue tests were conducted on a PLD-50 electro-hydraulic servo fatigue testing machine in laboratory static air at room temperature. Tests employed axial tension-compression fully reversed total strain control with a strain ratio R__ = -1. The applied nominal total strain amplitude ranged from 0.25% to 0.45%, with a triangular waveform at 1 Hz frequency. All fatigue tests were terminated when the cyclic stress amplitude dropped to 80% of its peak value during fatigue deformation, with the corresponding cycle count defined as the fatigue life under those conditions. At least two specimens were tested at each strain amplitude.

Precipitates and microstructures in the fatigue-deformed regions were examined using a TECNAI20 transmission electron microscope (TEM). TEM samples were prepared by electrolytic thinning of hand-ground 70 m thick foils using an MTP-1A twin-jet polisher with an electrolyte of 30 mL HNO + 70 mL CH OH at 15 V, 80–100 mA, and temperature -20 °C.

2.1 Microstructures of the Alloys

Figure 1 [Figure 1: see original paper] shows the microstructures of the T6-treated Al-9.0%Si-4.0%Cu-0.4%Mg(-0.3%Sc) alloys. The microstructures consist primarily of -Al matrix and spherical or oval eutectic Si phases. Comparison of Figs. 1a and 1b reveals that Sc addition effectively refines the microstructure of the Al-9.0%Si-4.0%Cu-0.4%Mg alloy.

Figure 2 [Figure 2: see original paper] presents TEM images and selected area electron diffraction patterns (SAEDP) of Al Sc phase in the Al-9.0%Si-4.0%Cu-0.4%Mg-0.3%Sc alloy before and after solution treatment. The morphology of Al Sc phase remains unchanged, indicating that the Al Sc precipitates formed during solidification do not dissolve during solution treatment.

Figure 3 [Figure 3: see original paper] shows TEM images and SAEDP of (Al Cu) phase in the T6-treated Al-9.0%Si-4.0%Cu-0.4%Mg(-0.3%Sc) alloys. Comparison of Figs. 3a and 3b demonstrates that the Sc-containing alloy contains a higher number density of phase, clearly showing that Sc addition promotes precipitation of phase.

2.2 Cyclic Stress Response Behavior

Figure 4 [Figure 4: see original paper] compares the cyclic stress response curves of T6-treated cast Al-9.0%Si-4.0%Cu-0.4%Mg and Al-9.0%Si-4.0%Cu-0.4%Mg-0.3%Sc alloys at given total strain amplitudes. At the same total strain amplitude, the cyclic stress amplitude of the T6-treated Al-9.0%Si-4.0%Cu-0.4%Mg-0.3%Sc alloy is significantly higher than that of the Sc-free alloy.

At a total strain amplitude Δ _t/2 = 0.25%, the Al-9.0%Si-4.0%Cu-0.4%Mg-0.3%Sc alloy initially exhibits cyclic strain hardening, with the hardening rate gradually decreasing as cycle number increases, eventually showing stable cyclic stress response in the mid-to-late stages of fatigue deformation. The Al-9.0%Si-4.0%Cu-0.4%Mg alloy, however, undergoes cyclic strain hardening throughout the entire fatigue deformation process, though the degree of hardening decreases with increasing cycles. At Δ _t/2 = 0.3%-0.45%, both alloys exhibit cyclic strain hardening throughout fatigue deformation, maintaining relatively high hardening rates at higher strain amplitudes without the decreasing hardening rate observed at lower amplitudes.

2.3 Low-Cycle Fatigue Life Behavior

Figure 5 [Figure 5: see original paper] shows the total strain amplitude versus low-cycle fatigue life curves for T6-treated Al-9.0%Si-4.0%Cu-0.4%Mg and Al-9.0%Si-4.0%Cu-0.4%Mg-0.3%Sc alloys. The low-cycle fatigue life of the Sc-containing alloy is higher than that of the Sc-free alloy at all strain amplitudes, demonstrating that Sc addition improves the low-cycle fatigue life of Al-9.0%Si-4.0%Cu-0.4%Mg alloy.

For total strain-controlled low-cycle fatigue tests, the relationship between plastic strain amplitude Δ _p/2, elastic strain amplitude Δ _e/2, and load reversal cycles to failure (2N_f) is typically described by [9]:

$$\frac{\Delta\varepsilon_t}{2} = \frac{\Delta\varepsilon_e}{2} + \frac{\Delta\varepsilon_p}{2} = \frac{\sigma_f'}{E}(2N_f)^b + \varepsilon_f'(2N_f)^c \tag{1}$$

where _f is the fatigue ductility coefficient, c is the fatigue ductility exponent, _f is the fatigue strength coefficient, b is the fatigue strength exponent, and E is Young' s modulus.

The cyclic stress-strain behavior can be described by the power law [10]:

$$\frac{\Delta\sigma}{2} = K'(\frac{\Delta\varepsilon_p}{2})^{n'} \tag{2}$$

where Δ /2 is the stress amplitude, K is the cyclic strength coefficient, and n is the cyclic strain hardening exponent.

Figure 6 [Figure 6: see original paper] shows the cyclic stress-strain curves of the T6-treated alloys. Figure 7 [Figure 7: see original paper] presents the strain amplitude versus load reversal cycles curves. The elastic strain amplitude shows a linear relationship with load reversal cycles, while the plastic strain amplitude exhibits a bilinear relationship: the slope in the high strain amplitude region is smaller than that in the low strain amplitude region. Using linear regression analysis of the experimental data, the strain fatigue parameters were determined and are listed in Table 1 .

Table 1 Strain fatigue parameters of T6-treated Al-9.0%Si-4.0%Cu-0.4%Mg(-0.3%Sc) alloys

Alloy	Δ _t/2 range	_f (MPa)	b	_f	c	K (MPa)	n
Al-Si-Cu-Mg	0.35%	[value]	[value	[value]	[value	[value]	[value]
Al-Si-Cu-Mg	0.35%	[value]	[value	[value]	[value	[value]	[value]
Al-Si-Cu-Mg-Sc	0.35%	[value]	[value	[value]	[value	[value]	[value]
Al-Si-Cu-Mg-Sc	0.35%	[value]	[value	[value]	[value	[value]	[value]

Notes: _f —fatigue strength coefficient; b—fatigue strength exponent; _f —fatigue ductility coefficient; c—fatigue ductility exponent; K —cyclic strength coefficient; n —cyclic strain hardening exponent

3. Analysis and Discussion

The microstructure refinement of Al-9.0%Si-4.0%Cu-0.4%Mg alloy by Sc addition can be explained as follows: (1) During solidification, Sc tends to enrich in the liquid phase ahead of the solid-liquid interface, increasing constitutional undercooling and promoting -Al nucleation [13]. Simultaneously, Sc enrichment at the solid-liquid interface increases local rare earth concentration, favoring Al Sc precipitation. Since Al Sc has an fcc structure with only 1.5% lattice mismatch with -Al, it exhibits good interfacial coherency with the -Al matrix. These pre-existing Al Sc particles can serve as effective heterogeneous nucleation sites for -Al during solidification, refining the as-cast grain structure [14]. With a melting point of 1320 °C [15], Al Sc does not dissolve during solution treatment and can effectively inhibit -Al grain growth during T6 processing. (2) During solidification, the surface of pre-solidified eutectic Si phases contains inherent steps that are twin-related to the -Al matrix, providing favorable sites for eutectic Si growth and resulting in the characteristic needle-like morphology of eutectic Si in as-cast Al-Si alloys [16]. Sc addition suppresses this preferential growth by adsorbing onto these inherent steps [17]. During solution treatment, eutectic Si morphology evolves through Si diffusion: initially, long needle-like Si particles undergo fission, and their sharp corners become blunted, transforming into several independent short rods. Subsequently, Si diffuses from the ends toward the middle of these rods, completing the spheroidization process to form spherical or oval Si particles. Generally, shorter needle-like eutectic Si in the as-cast structure leads to shorter rod-like Si after fission, reducing Si atom diffusion distances and resulting in higher spheroidization degree and smaller Si particle size.

The effect of Sc on precipitation in Al-9.0%Si-4.0%Cu-0.4%Mg alloy can be analyzed based on solute-vacancy interactions [18,19]. Sc has high binding energy with vacancies and strong solute-vacancy interactions. During solution treatment, vacancies gradually segregate around Sc atoms. After rapid quenching, these vacancy clusters collapse into dislocation loops. Cu atoms are attracted to the compressive stress fields of these dislocations during diffusion, and significant Cu segregation provides favorable nucleation sites for phase, accelerating its precipitation kinetics.

Cyclic strain hardening during fatigue deformation is associated with dislocation-dislocation interactions and dislocation-precipitate interactions. Figure 9 [Figure 9: see original paper] shows dislocation configurations in the Al-9.0%Si-4.0%Cu-0.4%Mg-0.3%Sc alloy after different cycles at Δ _t/2 = 0.25%. Initially, randomly distributed dislocations and dislocation tangles are observed (Fig. 9a). The rapid multiplication of dislocations and formation of tangles are considered the primary reason for high initial hardening rates. After 400 cycles (Fig. 9b), the dislocation structure evolves into parallel slip bands along a single direction, with some free-moving dislocations and few tangles between bands. Slip band formation allows rearrangement of the initially random dislocation network, producing some softening effect. Consequently, although cyclic hardening persists, the hardening rate is lower than in the initial stage. After 1000 cycles (Fig. 9c), with increasing cyclic stress amplitude, new slip systems activate when the applied load reaches the critical resolved shear stress, resulting in intersecting slip bands along different directions. Compared with Fig. 8c, no significant changes in dislocation configuration or density occur during continued deformation. Although the cyclic stress amplitude remains high, the hardening rate decreases markedly, even leading to stable cyclic stress response in the later stages.

At Δ _t/2 = 0.45% (Fig. 9d), numerous randomly distributed dislocations appear after just 3 cycles. With increasing cycles, extensive cellular substructures form (Fig. 8d). During cyclic deformation, dislocations primarily move in regions without cellular substructures and within cell interiors, causing some reduction in hardening rate. However, if cellular substructures form rapidly over most regions with high internal dislocation density or tangles, they strongly impede dislocation motion, maintaining high hardening rates [20]. The macroscopic cyclic stress response directly reflects changes in deformation resistance. Dislocation multiplication and tangle formation reduce dislocation mobility, increasing resistance to dislocation motion. Conversely, reconstruction of complex dislocation substructures into configurations with lower slip resistance produces softening effects. The overall cyclic stress response results from the competition between these hardening and softening mechanisms, with cyclic stability occurring when they reach equilibrium.

Sc addition significantly increases the cyclic stress amplitude of T6-treated Al-9.0%Si-4.0%Cu-0.4%Mg alloy for several reasons: (1) Sc refines the -Al grains in the as-cast alloy and inhibits their growth during T6 treatment, strengthening grain boundary strengthening; (2) Some Sc dissolves in the -Al matrix while some precipitates as Al Sc phase (Fig. 2), providing solid solution strengthening and second-phase strengthening; (3) Sc promotes dispersed precipitation of phase during aging, enhancing second-phase strengthening effects.

Microstructure affects crack propagation under fatigue loading [21]. In Al-Si cast aluminum alloys, the significant difference in mechanical properties between -Al matrix and eutectic Si phase causes Si particles to fail preferentially during fatigue deformation. Under applied loads, dislocations move primarily within the -Al matrix. When dislocations reach -Al/Si interfaces, they pile up and create stress concentrations. Larger eutectic Si particles are more prone to dislocation pile-up and stress concentration at interfaces. When stress concentration exceeds the fracture strength of Si particles or their bonding strength with the matrix, Si particle cracking or interface debonding occurs, leading to microcrack formation [22]. Sc addition reduces eutectic Si particle size in T6-treated Al-9.0%Si-4.0%Cu-0.4%Mg alloy, thereby decreasing stress concentration at interfaces and improving low-cycle fatigue life. Additionally, grain size affects fatigue crack propagation [23]. Grain boundaries directly impede crack growth. In Al-Si cast alloys, smaller -Al grains increase grain boundary density, raising the frequency with which fatigue cracks encounter grain boundaries during propagation. This enhanced barrier effect reduces crack growth rate and improves fatigue life.

The bilinear relationship between plastic strain amplitude and load reversal cycles in T6-treated Al-9.0%Si-4.0%Cu-0.4%Mg(-0.3%Sc) alloys has been observed in other materials and explained by differences in slip mechanisms, grain boundary rotation, microtwin formation, and changes in slip system activity [24–27]. For these alloys, the phenomenon can be interpreted by different plastic deformation mechanisms at high and low strain amplitudes [25]: At low strain amplitudes, plastic deformation is inhomogeneous, with dislocation slip concentrated in specific slip bands that pile up near grain boundaries, creating localized stress concentrations. At high strain amplitudes, dislocation slip is more uniform, allowing grains to accommodate deformation under higher applied loads. Thus, the change in plastic deformation mechanism with strain amplitude causes the slope change in the plastic strain amplitude versus load reversal cycles curve.

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4. Conclusions

1. At low total strain amplitudes, T6-treated Al-9.0%Si-4.0%Cu-0.4%Mg(-0.3%Sc) alloys exhibit cyclic strain hardening in the early stage of fatigue deformation, with hardening rate decreasing as cycle number increases, eventually showing stable cyclic stress response. At high total strain amplitudes, both alloys exhibit cyclic strain hardening throughout the fatigue deformation process.

2. Rare earth element Sc addition can simultaneously improve the cyclic deformation resistance and extend the low-cycle fatigue life of T6-treated Al-9.0%Si-4.0%Cu-0.4%Mg alloy.

3. Under low-cycle fatigue loading, the plastic deformation mechanism of T6-treated Al-9.0%Si-4.0%Cu-0.4%Mg(-0.3%Sc) alloys is closely related to the applied strain amplitude: planar slip occurs at low total strain amplitudes, while wavy slip dominates at high total strain amplitudes.

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Note: Figure translations are in progress. See original paper for figures.