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ARTICLE

# Influence of Ce Addition on Microstructure and Mechanical Properties of Al-Cu-Mn-Mg Alloy

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**Abstract:** This study systematically investigated the influence of cerium (Ce) addition on the microstructure and mechanical properties of Al-Cu-Mn-Mg alloys through multi-scale characterization techniques, including OM, SEM, TEM, and XRD. The results reveal that the addition of Ce leads to the formation of two distinct Ce-containing intermetallic compounds. One is the primary  $\text{Al}_{20}\text{Ti}_2\text{Ce}$  phase formed during solidification, which induces grain coarsening by consuming the effective grain-refining element Ti in the melt. The other Ce-containing phase is  $\text{Al}_{16}\text{Cu}_4\text{Mn}_2\text{Ce}$ , which forms through a eutectic reaction at the final stage of the solidification process. Unlike  $\text{Al}_2\text{Cu}$ , this phase is unstable at high temperatures and undergoes transformation into  $\text{Al}_{24}\text{Ce}_3\text{Cu}_8\text{Mn}$  and  $\text{Al}_{20}\text{Cu}_2\text{Mn}_3$  during heat treatment. Consequently, Cu and Mn from  $\text{Al}_{16}\text{Cu}_4\text{Mn}_2\text{Ce}$  phase cannot dissolve into the Al matrix, resulting in a reduction in supersaturation degree. Furthermore, the minor addition of Ce retards the phase transformation from  $\theta''$  to  $\theta'$  in the Al matrix. Thus, with the identical heat treatment, the primary strengthening phase is  $\theta''$  phase in the Ce-containing alloy, whereas the strengthening phase is  $\theta'$  phase in the alloy without Ce addition. These changes in the microstructure and phase composition directly influence the mechanical properties of alloys: with Ce content increasing from 0wt% to 0.3wt%, the ultimate tensile strength and yield strength are decreased by 95.0 and 45.5 MPa, respectively, and the elongation is reduced by about 50%, compared with those counterparts of the alloy without Ce addition.

**Key words:** Al-Cu-Mn-Mg alloy; cerium; microstructure; mechanical properties

## 1 Introduction

Al-Cu-based alloys are traditional heat-treatable materials used in the aerospace and automotive industries, owing to their properties of high strength, heat resistance, and low density. Micro-alloying represents a pivotal strategy for further enhancing the mechanical performance of these alloys<sup>[1-2]</sup>. In this context, rare earth (RE) elements have attracted considerable attention due to their profound influence on the microstructure and resultant properties of Al-Cu alloys<sup>[3-9]</sup>. The beneficial effects of RE additions are multifaceted, including

melt purification, grain refinement, and modification of phase composition and precipitation behavior.

Among various RE elements, Ce has been the subject of extensive research as a promising micro-alloying candidate for Al-Cu alloys. However, reported findings on its role appear complex and sometimes contradictory, heavily dependent on the specific alloy system. For instance, Song et al<sup>[10]</sup> reported that adding 0.45wt% Ce to an Al-Cu-Mg-Ag alloy markedly increased the nucleation density of the  $\Omega$  phase by providing heterogeneous nucleation sites while also enhancing the thermal stability of  $\Omega$  phase by impeding Cu

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atom diffusion. Li et al.<sup>[11]</sup> observed that introducing 0.25wt% Ce into a similar alloy led to the formation of  $\text{Al}_8\text{Cu}_4\text{Ce}$  phase during solidification, which resulted in a decrease in ultimate tensile strength and yield stress. Xiao<sup>[12]</sup> and Song<sup>[13]</sup> et al demonstrated that trace Ce addition could refine grain structure and promote a finer, denser distribution of  $\Omega$  and  $\theta'$  precipitates. However, they also noted that the co-addition of Ce and Ti resulted in the formation of  $\text{Al}_x\text{Ti}_6\text{Ce}_3\text{Cu}$  compounds, causing grain coarsening and reduction in tensile strength compared to alloys containing only Ce or Ti.

These conflicting reports underscore the fact that a definitive consensus regarding the role of Ce in Al-Cu alloys has yet to emerge, as its impact is intricately linked to the base composition<sup>[14–16]</sup>. Al-Cu-Mn-Mg alloys, as promising heat-resistant aluminum materials, have attracted considerable attention in recent years<sup>[2,17–18]</sup>. Therefore, this study systematically investigated the influence of Ce micro-alloying on an Al-Cu-Mn-Mg alloy. The objectives are to elucidate the underlying mechanisms governing microstructural evolution and to establish a clear correlation between Ce-induced microstructural changes and the resultant mechanical properties.

## 2 Experiment

The nominal composition (wt%) of the studied alloys are listed in Table 1. These alloys were fabricated using high-purity aluminum (99.99%), industrial-grade pure magnesium, Al-50Cu, Al-10Mn, Al-10Ce, and Al-5Ti-1B master alloys. All raw materials were thoroughly cleaned to remove surface contaminants such as dust, oils, moisture, and corrosive residues. High-purity aluminum, along with the Al-50Cu, Al-10Mn, and Al-10Ce intermediate alloys, was sequentially charged into a crucible placed inside a resistance furnace. The melt temperature was continuously monitored using a calibrated handheld thermometer. When the temperature of the melt reached  $700 \pm 5$  °C, industrial pure Mg was carefully introduced using a bell jar plunger, followed by stirring for 3–5 min to ensure homogeneous distribution of alloying elements. The melt temperature was then adjusted to 740–750 °C, after which the inoculant (Al-5Ti-1B) was added and manually stirred for 5–10 min. Subsequently, the temperature was adjusted to 730–740 °C, and the melt was refined using a non-toxic refining agent (HGJ-1A). After refining, the melt was held for 10–20 min to allow for slag separation, after which the surface slag was removed. Once the alloy melt temperature stabilized at 720 °C, the melt was poured into a preheated metal mold to obtain the experimental casting. T6 heat treatment was conducted in a muffle furnace. The samples were solution-treated at 525 °C for 12 h, water quenched, and then aged at 175 °C for 6 h to obtain a high number density of strengthening precipitates.

The phase composition of alloys was determined by X-ray diffraction (XRD, X'Pert PRO) with a scanning speed of  $10^\circ/\text{min}$  in the  $2\theta$  range of  $10^\circ$ – $90^\circ$ . Samples for microstructure analysis were prepared using standard metallographic methods. Microstructures were characterized using an optical microscope (OM, GX-71) and a scanning electron microscope

**Table 1 Chemical composition of experimental alloy (wt%)**

| Sample | Cu  | Mn  | Mg   | Ti  | Ce  | Al   |
|--------|-----|-----|------|-----|-----|------|
| 0Ce    | 5.0 | 1.0 | 0.35 | 0.3 | 0.0 | Bal. |
| 0.1Ce  | 5.0 | 1.0 | 0.35 | 0.3 | 0.1 | Bal. |
| 0.2Ce  | 5.0 | 1.0 | 0.35 | 0.3 | 0.2 | Bal. |
| 0.3Ce  | 5.0 | 1.0 | 0.35 | 0.3 | 0.3 | Bal. |

(SEM, FEI G3 UC). The chemical composition of secondary phases was analyzed using an energy-dispersive spectrometer (EDS) attached to the SEM and transmission electron microscope (TEM, JEM-2100F). Phase structures were further identified using selected area electron diffraction (SAED) analysis. Thin foils for TEM were prepared by cutting disks of 3 mm in thickness, grinding them to approximately 50  $\mu\text{m}$  in thickness, and final thinning by ion beam milling. Tensile tests were conducted on a WDW-50D microcomputer-controlled electronic universal testing machine at a crosshead speed of 2 mm/min. The Vickers hardness was measured using a microhardness tester (LM248AT, LECO, USA) under a load of 200 g ( $\text{HV}_{0.2}$ ) with a dwell time of 15 s. At least ten indentations were made for each condition, and the results were averaged.

## 3 Results and Discussion

### 3.1 Microstructure of as-cast alloys

Fig. 1 presents the microstructural evolution of Al-Cu-Mn-Mg alloys with varying Ce additions. The average grain size is quantified using ImageJ software. The results indicate that the average grain sizes are  $136 \pm 15$ ,  $214 \pm 24$ ,  $231 \pm 20$ , and  $326 \pm 48$   $\mu\text{m}$  for the 0Ce, 0.1Ce, 0.2Ce, and 0.3Ce alloys, respectively, demonstrating that Ce addition promotes grain coarsening. Concurrently, new secondary phases emerge with increasing Ce content. As shown in Fig. 1c<sub>1</sub>–Fig. 1d<sub>2</sub>, two distinct types of secondary phase particles are present in the 0.2Ce and 0.3Ce alloys. One is the light-gray blocky/needle-like phase (indicated by yellow arrows in Fig. 1) with well-defined geometric boundaries, predominantly distributed within  $\alpha$ -Al grains. The other is a black strip-like phase distributed at the grain boundaries (indicated by red arrows in Fig. 1). Additionally, the microstructure shown in Fig. 1b<sub>2</sub> indicates that black strip-like particles also form in the 0.1Ce alloy, and the light-gray secondary phase cannot be observed in this alloy. When the Ce content increases to 0.2wt%, gray blocky/needle-like particles begin to form within the  $\alpha$ -Al grain.

### 3.2 Phase composition

XRD patterns of the as-cast alloys with different Ce contents are presented in Fig. 2. All alloys consist primarily of  $\alpha$ -Al and  $\text{Al}_2\text{Cu}$  phases. For the high-Ce-containing (0.2Ce and 0.3Ce) alloys, an additional single diffraction peak appears at  $2\theta \approx 37.3^\circ$ , indicating the formation of a new secondary phase. However, due to its low volume fraction, this phase cannot be conclusively identified by XRD and is therefore marked as X in Fig. 2.

To identify the types of secondary phases observed by OM,

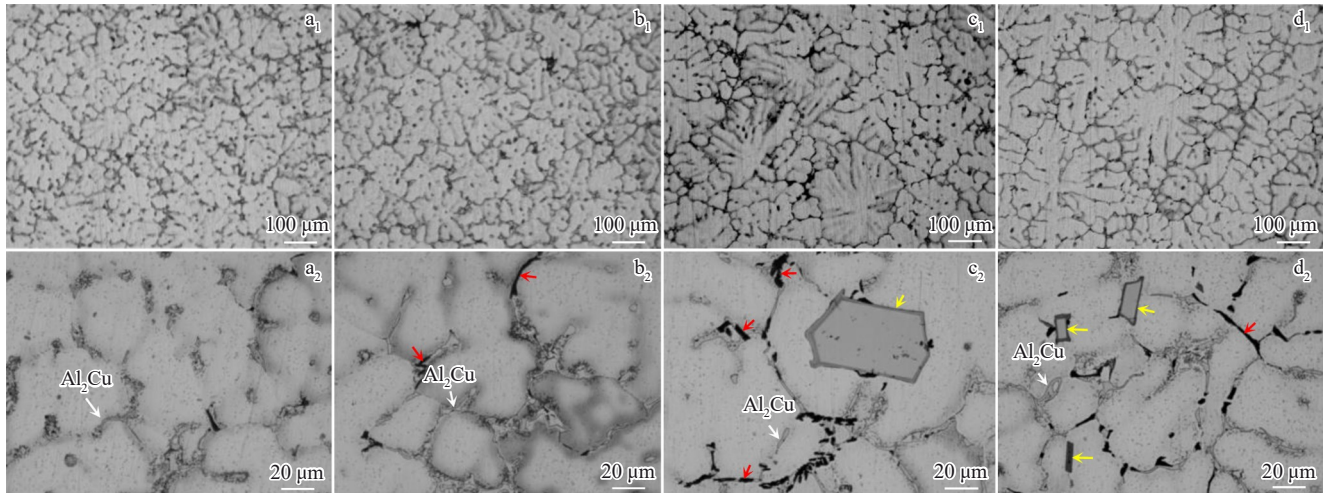


Fig.1 OM images of as-cast alloys: (a<sub>1</sub>–a<sub>2</sub>) 0Ce alloy; (b<sub>1</sub>–b<sub>2</sub>) 0.1Ce alloy; (c<sub>1</sub>–c<sub>2</sub>) 0.2Ce alloy; (d<sub>1</sub>–d<sub>2</sub>) 0.3Ce alloy

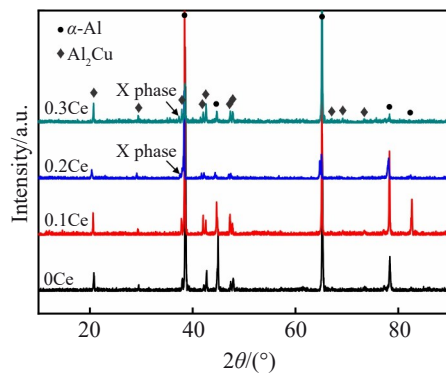


Fig.2 XRD patterns of as-cast Al-Cu-Mn-Mg alloys with different Ce contents

the 0.2Ce alloy and 0.3Ce alloy were further analyzed using SEM, TEM, and EDS. As shown in Fig.3, the phases within the grains exhibit polygonal and needle-like morphologies, which are characteristics of the primary phases formed during solidification, with sizes reaching about 40 and about 150  $\mu\text{m}$ , respectively. EDS analysis reveals that these block/needle-like phases are rich in Ti and Ce, along with minor amounts of Cu, Mg, and Mn. The atomic ratio of Al, Ti, and Ce is approximately 20: 2: 1, confirming the phase as  $\text{Al}_{20}\text{Ti}_2\text{Ce}$ . This compound forms by consuming Ti from the melt, which explains the observed grain coarsening (Fig. 1), as the Ti originally introduced for grain refinement is effectively depleted.

The secondary phases at the grain boundaries were also

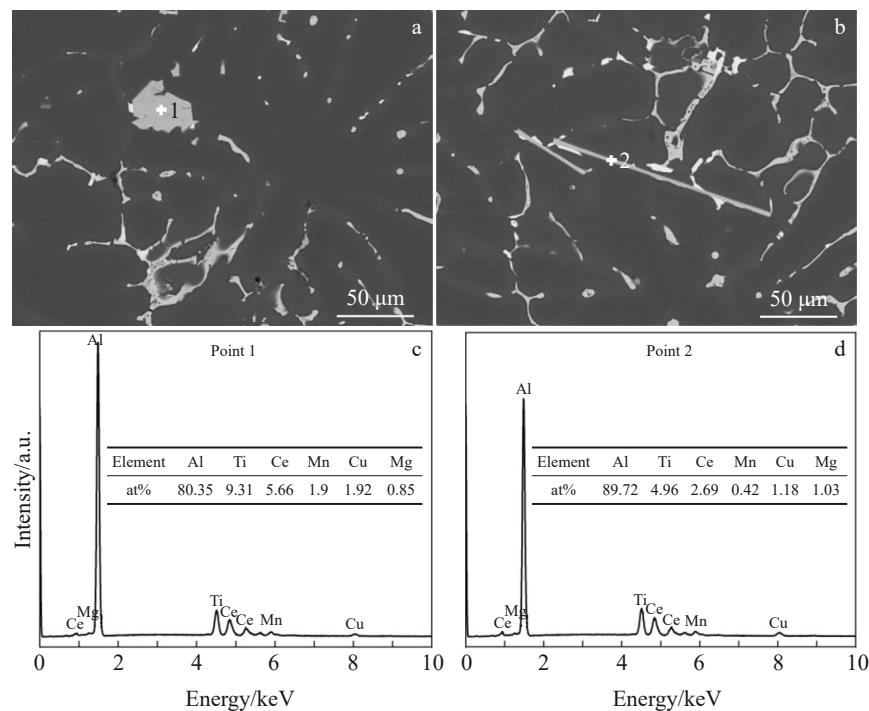


Fig.3 Typical secondary phase in the Al matrix of as-cast 0.3Ce alloy: (a–b) SEM images at back-scattered electron mode; (c–d) EDS analysis results

characterized by SEM/TEM/EDS. As shown in Fig.4a, a phase with brighter contrast than that of  $\text{Al}_2\text{Cu}$  is distributed along the boundaries. EDS analysis (Fig.4b–4c) confirms that this phase is distinct from the intragranular  $\text{Al}_{20}\text{Ti}_2\text{Ce}$  and can be identified as  $\text{Al}_{16}\text{Cu}_4\text{Mn}_2\text{Ce}$ , consistent with prior results<sup>[19]</sup>. These results indicate that the solidification of the Ce-containing alloy involves a complex eutectic reaction:  $\text{L} \rightarrow \text{Al} + \text{Al}_2\text{Cu} + \text{Al}_{16}\text{Cu}_4\text{Mn}_2\text{Ce}$ .

The grain boundary phase was further analyzed by TEM/EDS. As shown in Fig.5b, EDS result of the circled region in Fig.5a indicates that the strip-like particle is  $\text{Al}_{16}\text{Cu}_4\text{Mn}_2\text{Ce}$ . The corresponding SAED patterns (Fig.5c–5d) were acquired from this phase. The hexagonal symmetry of the diffraction spots in Fig.5c is consistent with a face-centered cubic (fcc), hexagonal close-packed (hcp), or body-centered cubic (bcc) structure. Indexing of both patterns confirms an fcc structure for  $\text{Al}_{16}\text{Cu}_4\text{Mn}_2\text{Ce}$ , with the patterns corresponding to the

$[\bar{1}11]$  and  $[\bar{2}23]$  zone axes.

3.3 Phase transformation

Fig. 6 shows the microstructure and corresponding EDS analysis of the heat-treated 0.3Ce alloy. The results confirm that the blocky/needle-like phase is still  $\text{Al}_{20}\text{Ti}_2\text{Ce}$ , whose morphology and composition show negligible alteration after heat treatment. This observation is consistent with previous results<sup>[19]</sup>, further confirming the exceptionally high thermal stability of this intermetallic phase, and it also shows almost no dissolution and negligible morphological change during the solution treatment at 525 °C.

Fig.7a shows the secondary phases at grain boundary in the heat-treated 0.3Ce alloy. Unlike the strip-like  $\text{Al}_{16}\text{Cu}_4\text{Mn}_2\text{Ce}$  observed in the as-cast alloy, these particles exhibit a distinct reticular morphology. However, EDS analysis results (Fig.7b–7c) confirm that their composition remains consistent with

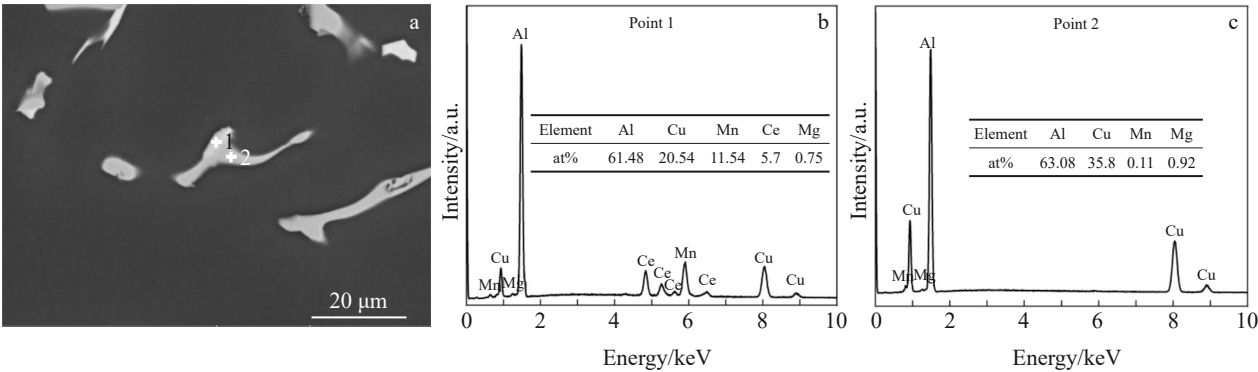


Fig.4 Typical secondary phase at the grain boundary of as-cast 0.3Ce alloy: (a) SEM image at back-scattered electron mode; (b–c) EDS analysis results

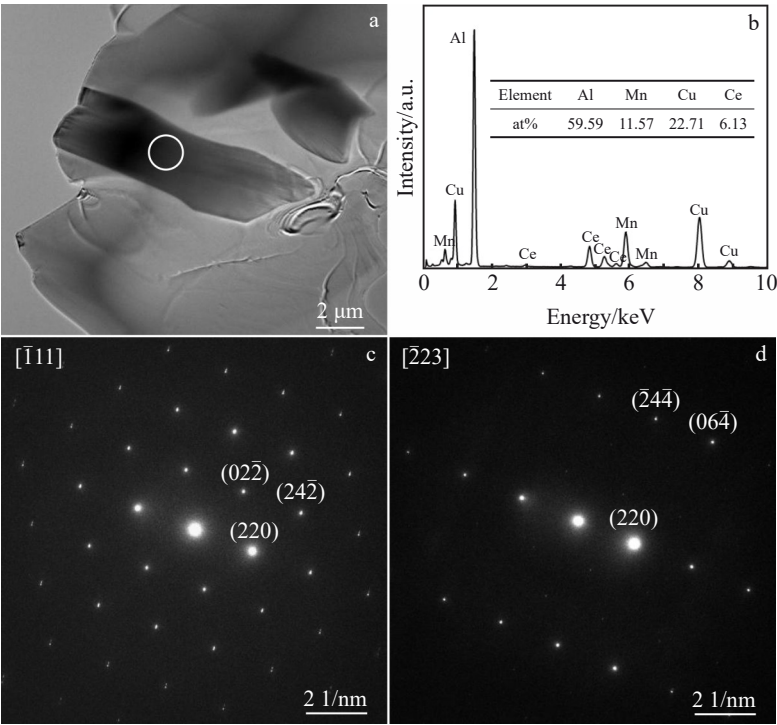


Fig.5 Strip-like phase in the as-cast 0.3Ce alloy: (a) bright-field TEM image; (b) EDS analysis results; (c–d) SAED patterns

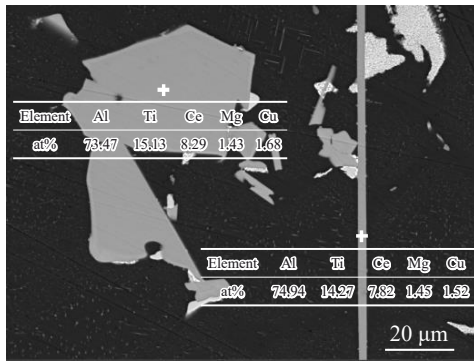


Fig.6 SEM image at back-scattered electron mode with EDS analysis results of block/needle-like phase in heat-treated 0.3Ce alloy

$\text{Al}_{16}\text{Cu}_4\text{Mn}_2\text{Ce}$ . Moreover, the internal contrast variations within the particles are evident. These features collectively suggest that the  $\text{Al}_{16}\text{Cu}_4\text{Mn}_2\text{Ce}$  phase has undergone a transformation during heat treatment, despite retaining its

nominal chemical composition.

TEM analysis was employed to investigate the phase transformation of  $\text{Al}_{16}\text{Cu}_4\text{Mn}_2\text{Ce}$  during heat treatment (Fig.8). EDS mapping reveals two distinct phases within the observed region. In the high-angle annular dark field image, the bright particles are enriched in Ce and Cu, along with some Mg and Mn. In contrast, the dark particles are rich in Mn and Cu but contain no Ce.

The two phases were analyzed by SAED and EDS, with results shown in Fig.9. Based on the combined SAED patterns and EDS compositional measurements, the bright and dark particles are identified as  $\text{Al}_{24}\text{Ce}_3\text{Cu}_8\text{Mn}$  and  $\text{Al}_{20}\text{Cu}_2\text{Mn}_3$  (T phase), respectively. This confirms that the  $\text{Al}_{16}\text{Cu}_4\text{Mn}_2\text{Ce}$  phase formed during solidification is metastable and transformed into  $\text{Al}_{24}\text{Ce}_3\text{Cu}_8\text{Mn}$  and T phases during heat treatment.

### 3.4 Mechanical properties

Fig.10 shows the engineering stress-engineering strain curves of Al-Cu-Mn-Mg alloy with different Ce contents at room

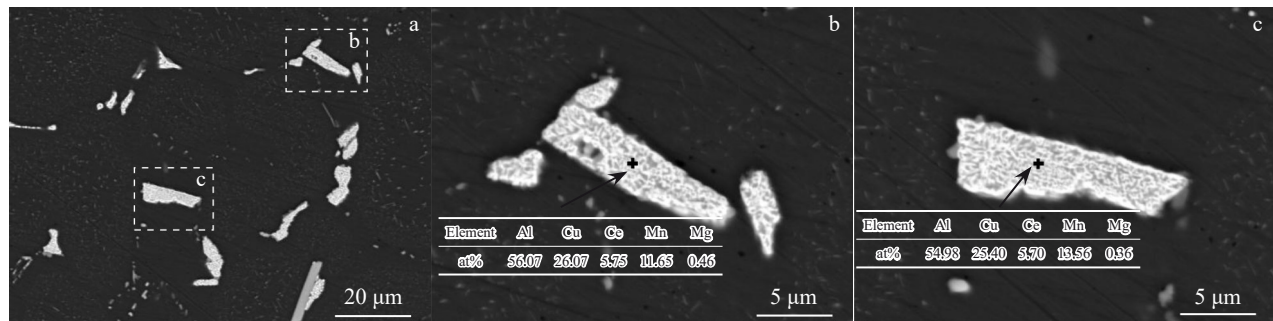


Fig.7 Secondary phases at the grain boundaries in heat-treated 0.3Ce alloy: (a) SEM image at back-scattered electron mode; (b–c) enlarged images of the selected areas in Fig.7a with EDS analysis results

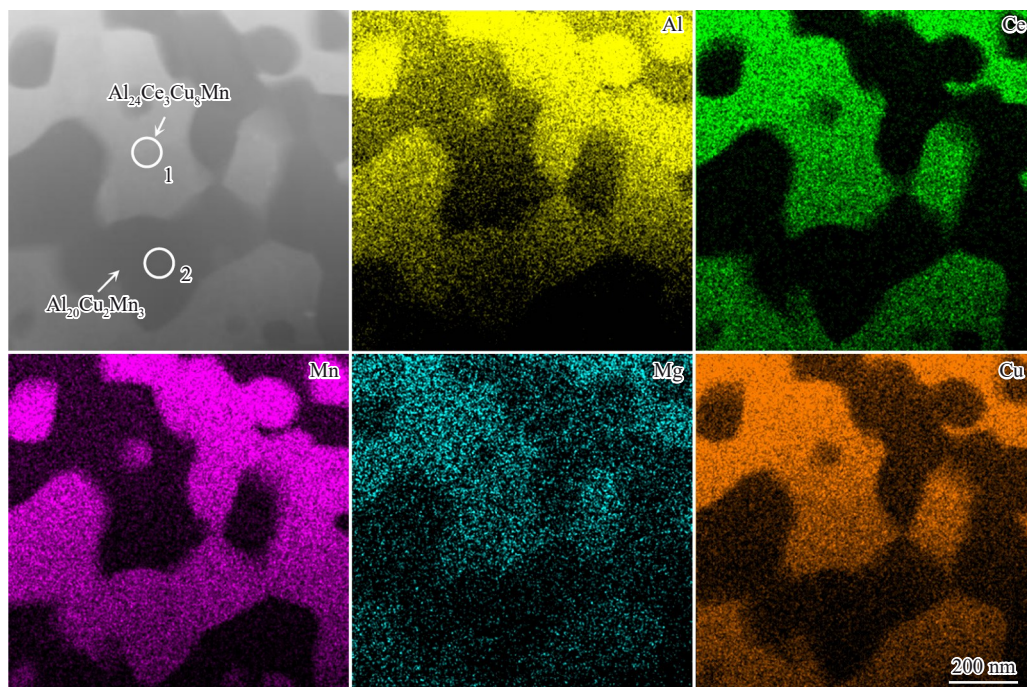


Fig.8 SEM image and corresponding EDS mapping results of the secondary phase in heat-treated 0.3Ce alloy

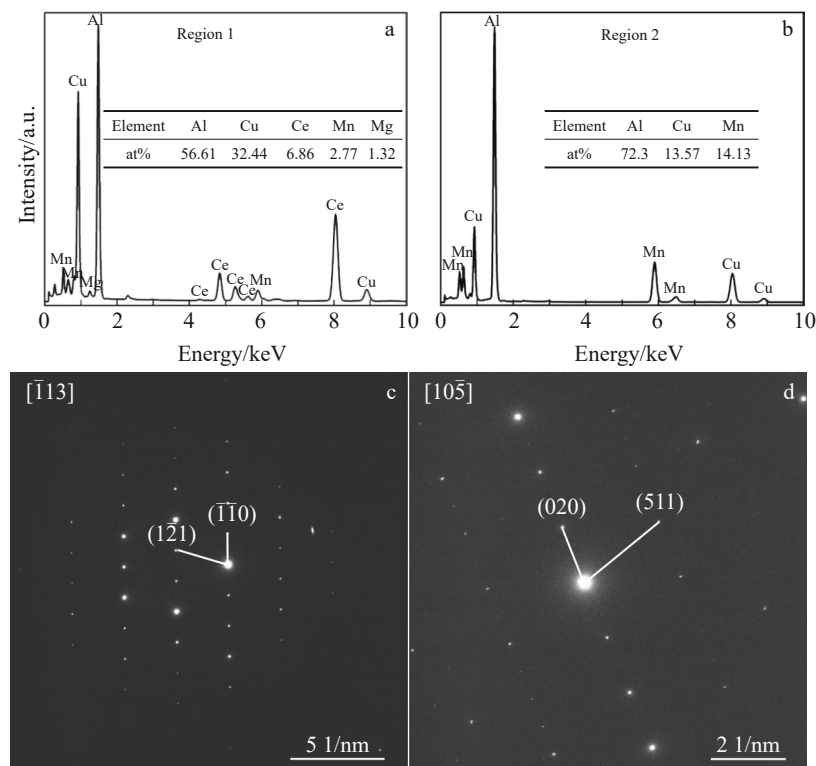


Fig.9 EDS analysis results (a–b) and SAED patterns (c–d) of the region 1 (a, c) and region 2 (b, d) marked in Fig.8

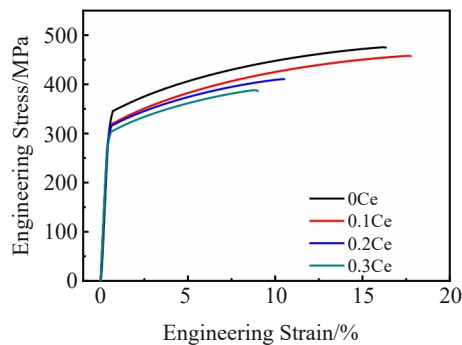


Fig.10 Engineering stress-engineering strain curves of Al-Cu-Mn-Mg alloy with different Ce contents

temperature. The mechanical properties are listed in Table 2. Ce micro-alloying significantly degrades both the strength and plasticity. Specifically, with Ce content increasing from 0wt% to 0.3wt%, the yield strength (YS), ultimate tensile strength (UTS), and elongation (EL) basically decrease from 343.0±3.0 MPa, 468.0±7.0 MPa, and 7.0%±1.0% to 297.5±6.5 MPa, 373.0±15.0 MPa, and 3.0%±1.0%, respectively. With the

Table 2 Mechanical results of Al-Cu-Mn-Mg alloy with different Ce contents at room temperature

| Alloy | UTS/MPa    | YS/MPa    | EL/%     |
|-------|------------|-----------|----------|
| 0Ce   | 468.0±7.0  | 343.0±3.0 | 7.0±1.0  |
| 0.1Ce | 453.0±5.0  | 319.5±1.5 | 8.0±10.0 |
| 0.2Ce | 408.0±3.0  | 311.0±4.0 | 4.5±2.5  |
| 0.3Ce | 373.0±15.0 | 297.5±6.5 | 3.0±1.0  |

lower Ce content (0.1wt%), UTS shows a moderate decrease, while the YS drops dramatically, indicating a reduced strengthening contribution from minor Ce addition. According to the analysis, this is partly due to the formation of Ce-containing compounds ( $\text{Al}_{16}\text{Cu}_4\text{Mn}_2\text{Ce}$  in the as-cast state, and  $\text{Al}_{24}\text{Ce}_3\text{Cu}_8\text{Mn}$  and  $\text{Al}_{20}\text{Cu}_2\text{Mn}_3$  after heat treatment), which reduces the supersaturation of Cu and Mn near the grain boundaries. This depletion impairs the precipitation of strengthening phases during heat treatment, resulting in a lower precipitate density around the eutectic regions and consequently a decrease in YS compared to the Ce-free alloy. As the Ce content increases from 0.1wt% to 0.2wt%, the YS remains nearly constant, suggesting that the effect of Ce alloying on strengthening tends to stabilize. Besides, UTS and YS decrease by 95.0 and 45.5 MPa, respectively, and EL is reduced by about 50%, compared to those properties of the alloy without Ce addition.

The tensile fracture surfaces of the alloys with different Ce contents are shown in Fig.11. The Ce-free (0Ce) alloy exhibits numerous fine ductile dimples, indicative of substantial plastic deformation. With the addition of 0.1wt% Ce, the area fraction of ductile dimples decreases. This is because the eutectic at grain boundary ( $\text{Al}_{24}\text{Ce}_3\text{Cu}_8\text{Mn}$  and  $\text{Al}_{20}\text{Cu}_2\text{Mn}_3$ ) formed during heat treatment provides preferential sites for fracture initiation. Notably, no  $\text{Al}_{20}\text{Ti}_2\text{Ce}$  particles are observed on this fracture surface. When the Ce content reaches 0.2wt%, blocky  $\text{Al}_{20}\text{Ti}_2\text{Ce}$  particles become visible on the fracture surface, accompanied by clear cleavage facets. As  $\text{Al}_{20}\text{Ti}_2\text{Ce}$  is a brittle intermetallic compound, it readily initiates cracks during tensile testing, which accounts for the continued decrease in

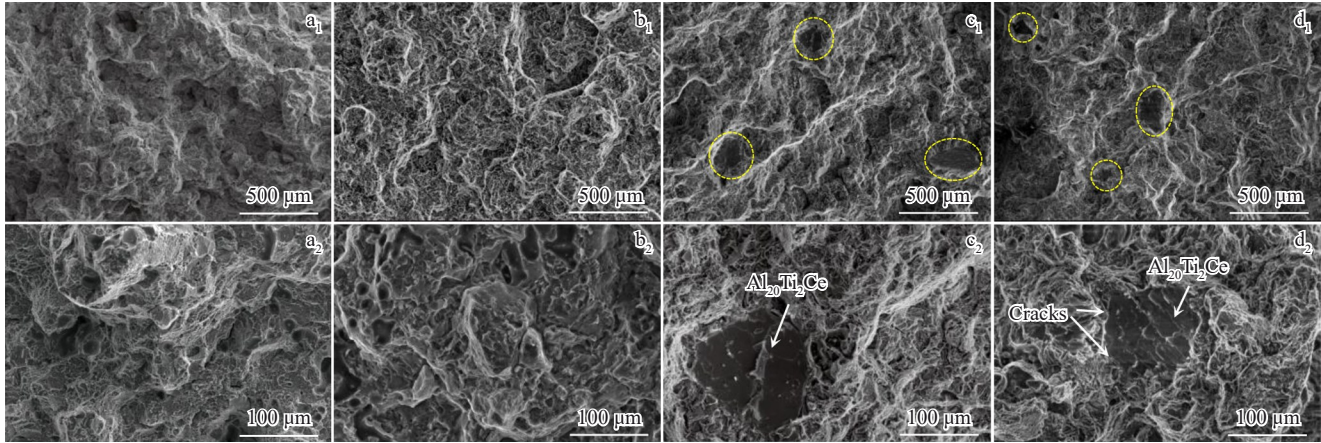


Fig.11 SEM images at secondary electron mode of the tensile fracture surfaces: (a<sub>1</sub>–a<sub>2</sub>) 0Ce alloy; (b<sub>1</sub>–b<sub>2</sub>) 0.1Ce alloy; (c<sub>1</sub>–c<sub>2</sub>) 0.2Ce alloy; (d<sub>1</sub>–d<sub>2</sub>) 0.3Ce alloy

elongation with increasing Ce content.

To elucidate the marked decline in YS, the key strengthening precipitates in the alloys were characterized. In precipitation-hardened Al-Cu alloys, room-temperature strength is predominantly governed by the characteristics of age-hardening precipitates, including their type, size, and volume fraction. The differences in mechanical strength among the experimental alloys (Fig. 10) can be effectively interpreted using an established strength model<sup>[20]</sup>.

The 0.2Ce alloy is taken as representative of the Ce-containing alloys. Bright-field TEM images and corresponding SAED patterns of the heat-treated 0Ce and 0.2Ce alloys are shown in Fig.12. The microstructure of the 0Ce alloy is dominated by  $\theta'$  precipitates<sup>[21]</sup>, as confirmed by the characteristic diffraction spots in Fig.12b. In contrast, the SAED pattern of the 0.2Ce alloy (Fig.12d) exhibits discontinuous streaks along (200)<sub>Al</sub> with intensity maxima at the 1/4, 1/2, and 3/4 positions, identifying  $\theta''$  as the primary precipitate. These results demonstrate that the addition of minor Ce retards the  $\theta'' \rightarrow \theta'$  transformation, leading to different dominant precipitates in the two alloys despite identical heat treatment.

To quantify the strengthening contribution of  $\theta''/\theta'$  precipitates, their diameter and thickness distributions were statistically analyzed, as shown in Fig. 13. The 0.2Ce alloy has significantly finer precipitates, with an average diameter of  $33.2 \pm 1.3$  nm

and thickness of  $1.799 \pm 0.044$  nm. In contrast, the precipitates in the 0Ce alloy are considerably coarser, averaging  $109.9 \pm 5.1$  nm in diameter and  $4.679 \pm 0.272$  nm in thickness. Additional quantitative TEM data are summarized in Table 3.

The  $\theta''$  phase enhances alloy strength through multiple mechanisms including interfacial strengthening, coherency strengthening, and modulus mismatch strengthening, all of which originate from the interactions between the precipitates and dislocations<sup>[22–23]</sup>. It is well established that a higher volume fraction of  $\theta''$  precipitates leads to a significant increase in strength. Due to their coherency with the  $\alpha$ -Al matrix and their shearable nature,  $\theta''$  precipitates contribute primarily through coherency strengthening and interfacial strengthening. In contrast, the effects of modulus mismatch strengthening and order strengthening are generally considered negligible<sup>[22–24]</sup>. The contribution of the  $\theta''$  precipitates to the critically resolved shear stress (CRSS)  $\Delta\tau$  due to coherency strengthening can be estimated<sup>[25]</sup> as follows.

$$\Delta\tau = 4.1G|\epsilon|^{3/2}\left(\frac{fd_l}{2b}\right)^{1/2} \quad (1)$$

where  $G$  is the shear modulus of the  $\alpha$ -Al matrix (28 GPa<sup>[25]</sup>),  $\epsilon$  is the lattice strain (0.006<sup>[24]</sup>),  $f$  is the volume fraction of precipitates,  $b$  is the magnitude of the Burgers vector for aluminum (0.284 nm<sup>[24]</sup>), and  $d_l$  is the effective width of the  $\theta''$

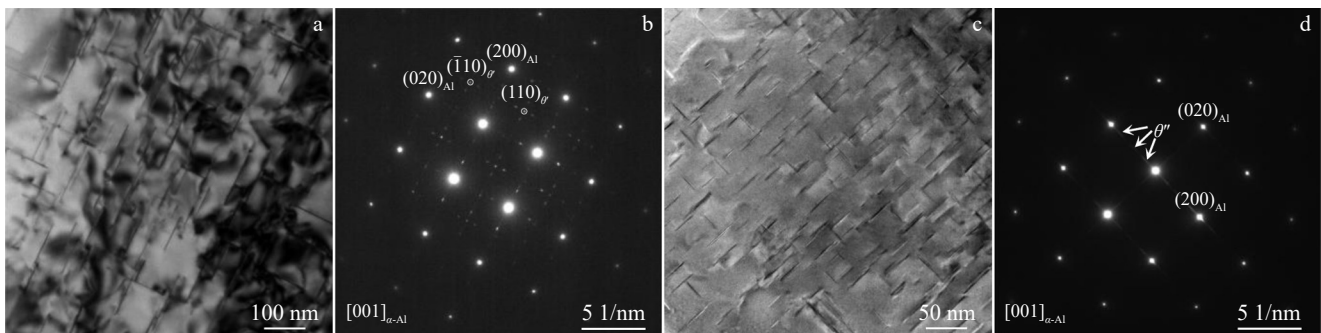


Fig.12 Bright-field TEM images (a, c) and corresponding SAED patterns (b, d) of different alloys along  $\langle 001 \rangle_{\alpha\text{-Al}}$ : (a–b) 0Ce alloy; (c–d) 0.2Ce alloy

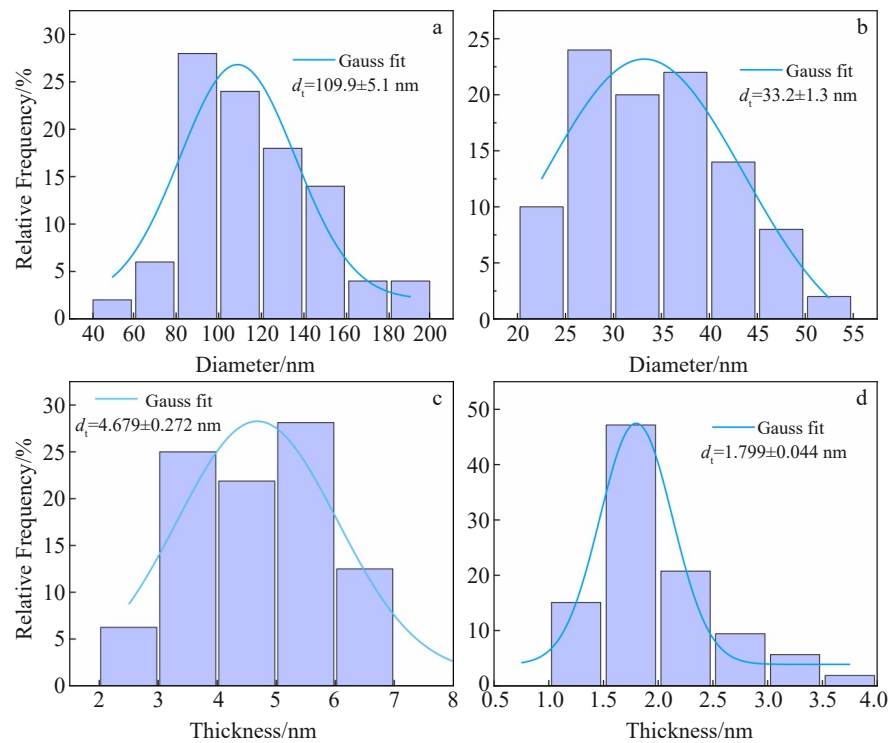


Fig.13 Distribution of diameter (a–b) and thickness (c–d) of  $\theta'/\theta''$  phases in 0Ce (a, c) and 0.2 Ce (b, d) alloys at T6 state

precipitates.

The increase in the CRSS due to interfacial strengthening of the  $\theta''$  phase results from the formation of new interfaces during its interactions with dislocations<sup>[22,24]</sup> and can be estimated as<sup>[22]</sup>:

$$\Delta\tau = \left( \frac{0.908d_t}{t_i^2} \right) \left( \frac{bf}{\Gamma} \right)^{\frac{1}{2}} \gamma_i^{\frac{3}{2}} \tag{2}$$

where  $t_i$  is the effective thickness of the  $\theta''$  precipitates;  $\gamma_i$  is the interfacial energy (0.21 J/m<sup>2</sup><sup>[17]</sup>), and  $\Gamma$  represents the dislocation line tension.  $\Gamma$  is expressed as:

$$\Gamma = \frac{Gb^2}{2\pi} \ln \sqrt{\frac{d_t^2}{2b^2f}} \tag{3}$$

The CRSS increments due to coherency and interfacial strengthening by the  $\theta''$  precipitates in the 0.2Ce alloy are calculated as 50.6 and 38.8 MPa, respectively. The YS increment due to precipitates  $\sigma_p$  is obtained as:

$$\sigma_p = M\Delta\tau \tag{4}$$

where  $M$  is the Taylor factor ( $M=3$ <sup>[24]</sup>). As shown in Table 3, the  $\sigma_p$  value of the 0.2Ce alloy is estimated to be 268.2 MPa, which accounts for about 86% of the experimentally measured YS. This suggests reasonable agreement between the estimated YS increment due to  $\theta''$  precipitates and the measured YS, because the strengthening effects of other mechanisms, such

as solid solution strengthening and grain size hardening in the T6-treated alloys, remain relatively slight.

Furthermore, the  $\theta'$  particles formed in the 0Ce alloy are non-shearable; therefore, they contribute to alloy strengthening via the Orowan looping mechanism<sup>[24]</sup>. The CRSS increment due to  $\theta'$  particles with the observed characteristics (Table 3) is estimated<sup>[20]</sup> as:

$$\Delta\tau = \left( \frac{Gb}{2\pi\sqrt{1-\nu}} \right) \left( \frac{1}{1.23\frac{1.03}{\sqrt{N_vd_t}} - \frac{\pi d_t}{8} - 1.061t_i} \right) \times \left( \ln \frac{0.981\sqrt{d_t t_i}}{b} \right) \tag{5}$$

where  $N_v$  is the number density of  $\theta'$  precipitates, and  $\nu$  is the Poisson ratio (0.33 for fcc cubic metals<sup>[20]</sup>). The YS increments due to  $\theta'$  particles in the 0Ce alloy are estimated to be 325.2 MPa, which confirms that the  $\theta'$  precipitates observed in the 0Ce alloy make a much larger contribution to the strengthening of the alloy than  $\theta''$  precipitates observed in the 0.2Ce alloy (Table 2). Therefore, the reduction in YS shown in Fig. 10 can be primarily attributed to the modification of the strengthening phase induced by the addition of Ce. In other words, under identical heat treatment conditions, the 0Ce alloy reaches peak-

Table 3 Characteristics of  $\theta''$  and  $\theta'$  phases in 0Ce and 0.2Ce alloys at T6 state

| Alloy | Phase      | Mean diameter, $d_t$ /nm | Mean thickness, $t_i$ /nm | Number density, $N_v/\mu\text{m}^{-3}$ | Volume fraction, $f/\%$ | $\Delta\tau$ /MPa | $\sigma_p$ /MPa |
|-------|------------|--------------------------|---------------------------|----------------------------------------|-------------------------|-------------------|-----------------|
| 0Ce   | $\theta'$  | 109.9±5.1                | 4.679±0.272               | 1200.0                                 | 0.0532                  | 108.4             | 325.2           |
| 0.2Ce | $\theta''$ | 33.2±1.3                 | 1.799±0.044               | 9877.6                                 | 0.0154                  | 89.4              | 268.2           |

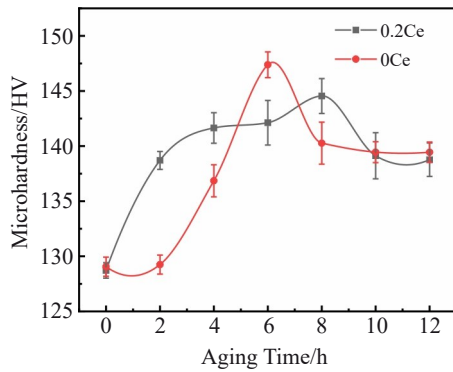


Fig.14 Microhardness-aging time curves for 0Ce and 0.2Ce alloys aged at 175 °C

aged hardness at 175 °C for 6 h, and the Ce-containing alloys cannot reach the peak hardness, because the addition of Ce can retard the transformation from  $\theta''$  to  $\theta'$ .

To verify that Ce retards the  $\theta'' \rightarrow \theta'$  transformation, the microhardness-aging time curves for the 0Ce and 0.2Ce alloys are plotted in Fig. 14 (microhardness was measured in the  $\alpha$ -Al matrix). The peak hardness occurs at 6 h for the 0Ce alloy and at 8 h for the 0.2Ce alloy. Moreover, the 0Ce alloy exhibits a higher peak-aged microhardness ( $147.4 \pm 1.2$  HV) than the 0.2Ce alloy ( $144.5 \pm 1.7$  HV).

Furthermore, the effect of grain size on the mechanical properties of the 0Ce and 0.2Ce alloys has also been analyzed.  $\sigma_p$  is dependent on the average grain boundary diameter ( $D$ ) by the Hall-Petch relationship.

$$\sigma_p = \sigma_0 + kD^{-1/2} \quad (6)$$

where  $\sigma_0$  is the intrinsic resistance of the lattice to dislocation motion, and  $k$  is the constant representing the relative strengthening contribution from grain boundaries. For Al alloys,  $k$  is usually taken to be  $0.14 \text{ MN/m}^{3/2}$ . According to the Hall-Petch equation, the theoretical decrease in  $\sigma_p$  due to grain coarsening from  $136 \mu\text{m}$  to  $231 \mu\text{m}$  is only about  $2.8 \text{ MPa}$ <sup>[27]</sup>. This suggests that the influence of grain size variation on the mechanical properties of alloys is negligible.

## 4 Conclusions

1) Primary  $\text{Al}_{20}\text{Ti}_2\text{Ce}$  and eutectic  $\text{Al}_{16}\text{Cu}_4\text{Mn}_2\text{Ce}$  phases can be formed during solidification, when the Ce element is added to an Al-Cu-Mn-Mg alloy. The volume fraction of these Ce-containing quaternary compounds increases with increasing Ce content.

2)  $\text{Al}_{20}\text{Ti}_2\text{Ce}$  is highly stable, and its morphology remains unchanged during heat treatment. In contrast,  $\text{Al}_{16}\text{Cu}_4\text{Mn}_2\text{Ce}$  is unstable; this eutectic phase is transformed into  $\text{Al}_{24}\text{Ce}_3\text{Cu}_8\text{Mn}$  and  $\text{Al}_{20}\text{Cu}_2\text{Mn}_3$  during heat treatment.

3) The addition of minor Ce retards the phase transformation from  $\theta''$  to  $\theta'$  in the Al matrix. As the Ce content increases from 0wt% to 0.3wt%, the mechanical properties of the alloys deteriorate: UTS and YS decrease by 95.0 and 45.5 MPa, respectively, and EL decreases by about 50%, compared with those of the Ce-free base alloy.

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## 稀土元素Ce对Al-Cu-Mn-Mg合金显微组织及力学性能的影响

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**摘 要:** 采用OM、SEM、TEM及XRD等表征技术研究了微量Ce对Al-Cu-Mn-Mg合金显微组织与力学性能的影响。结果表明: 随着Ce含量的增加, 合金中会形成两种不同的含Ce化合物, 一种是凝固过程中形成的初生 $\text{Al}_{20}\text{Ti}_2\text{Ce}$ 相, 该相的出现消耗了Al熔体中有效的晶粒细化元素Ti, 导致合金的晶粒粗化。另一种是凝固后期通过共晶反应形成的 $\text{Al}_{16}\text{Cu}_4\text{Mn}_2\text{Ce}$ 相, 该相在高温下不稳定, 热处理过程中会转变为 $\text{Al}_{24}\text{Ce}_3\text{Cu}_8\text{Mn}$ 和 $\text{Al}_{20}\text{Cu}_2\text{Mn}_3$ , 无法使 $\text{Al}_{16}\text{Cu}_4\text{Mn}_2\text{Ce}$ 相中的Cu、Mn元素固溶入Al基体, 导致热处理后合金基体中Cu、Mn元素的过饱和度下降。此外, Ce的添加会延缓Al基体中 $\theta''$ 相向 $\theta'$ 相的转变过程, 同等热处理条件下, 含Ce合金的Al基体中的主要强化相为 $\theta''$ , 而不含Ce合金中的强化相为 $\theta'$ 。稀土元素Ce对合金相组成及显微组织的影响直接体现在合金力学性能的变化上, 当Ce含量由0wt%增加至0.3wt%时, 与不含Ce的基础合金相比, 含0.3wt% Ce合金的极限抗拉强度和屈服强度分别下降了95.0和45.5 MPa, 伸长率下降了约50%。

**关键词:** Al-Cu-Mn-Mg合金; Ce; 显微组织; 力学性能

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