Alloying effect of Mg on microstructure and mechanical properties at 300 °C of Al–5Cu–1Mn–0.5Ni heat-resistant alloy

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Abstract

The effect of Mg alloying on microstructure and mechanical properties at 300 °C of Al–5Cu–1Mn–0.5Ni heat-resistant aluminum alloy was investigated by microstructure observation and tensile test. Alloying with 0.5 wt%Mg addition considerably increases the yield strength at 300 °C of samples at as-cast, as-solutionized and as-aged states. It also greatly enhances the strengthening effects from solution and ageing treatments, however, simultaneously, decreases the elongation greatly. During the solidification process of the alloy with Mg addition, a great amount of whisker-like Al–Cu–Mg compound phase is formed at the eutectic boundary zones in the form of ridge configuration, which is determined as AlCuMg intermetallic compound by TEM. During the solution treatment, this whisker-like AlCuMg phase is completely dissolved into Al matrix that leads to a remarkable rise in the increment of yield strength at 300 °C by solution strengthening. The alloying of Mg has a significant influence on the precipitation behavior during the ageing treatment, for examples, promoting the formation of finer θ' particles and a great number of Cu–Mg atomic clusters. During high temperature tensile test, theses clusters evolve into fine cube-like Al₅Cu₂Mg particles with orthorhombic structure. It is the significant microstructure evolution induced by alloying of Mg that leads to the great rise in yield strength at 300 °C. Addition of Mg in Al–Cu–Mn–Ni alloy dose not lead to the formation of meta-stable or stable S phase, but whisker-like AlCuMg and cube-like Al₅Cu₂Mg compounds are determined by TEM.

1. Introduction

2219 and 2519 aluminum alloys (Al–Cu–Mn system) have been applied in the aircraft industry due to its high mechanical properties at elevated temperature (up to 250 °C) [1]. However, with the increasing requirement in the aircraft flight velocity, conventional Al-Cu-Mn alloys cannot meet the demands for the more rigorous working environment, because θ' meta-stable phase of Al₂Cu, the main strengthening particles in this alloy system, starts to coarsen quickly when the temperature is over 250 °C[2] and transforms into coarse Al₂Cu phase [3], losing the strengthening effect.

In this alloy system, the addition of Ni, Fe and RE elements, and some trace elements like Ti, Zr, B, V, Mg, Er and Sc, etc can enhance the elevated temperature strength to some extent. Zhao et al [4] reported that 1.5% La/Ce addition in Al–5.0Cu–0.6Mn alloy could improve the high temperature tensile strength to 102.4 MPa at 300 °C, owing to the formation of heat-resistant phases during the solution treatment. Lin et al [5, 6] reported that 0.17% Yb addition in 2519 alloy improved the high temperature tensile strength at 300 °C from 139.5 MPa to 169.4 MPa, because the density of θ' phase was increased. Wang et al [6] thought that the improvement in the high temperature tensile strength of 2519 A alloy by 0.4%Ce addition was due to the formation of high density θ' precipitates and high melting point Al₅Cu₂Ce phase. Lin et al [7] reported that the mechanical properties of
squeezed casting Al–5.0Cu–0.6Mn–1.0Fe alloy at 300 °C was increased with Ni content from 0.5% to 1.5%, owing to the increased amount of the heat-resistant Al9FeNi and Al3CuNi phases. Chen et al. [8] thought that small dispersed particles formed during the solutionizing course and fine precipitates formed during the ageing course played a much greater role in high temperature strengthening. Especially, when 0.5 wt% Ni is added into Al–5Cu–1Mn alloy, the ageing treatment produces a significant increment in yield strength (YS) at 300 °C by 26 MPa. The increment in YS at 300 °C from the ageing treatment is much larger than that from the Ni-rich heat resistant phases formed during solidification (12 MPa induced by 1.5 wt% Ni addition). The development of heat resistant aluminum alloys should pay more attention to enhancing the precipitation during solution and ageing treatments and to improve the thermal stability of the precipitation phases.

Mg is also often added into Al–Cu alloy to improve mechanical properties, because it promotes the formation of S–Al13Cu5Mg phase [9]. S phase is thought to be a strengthening phase like θ-Al13Cu5 phase. During the ageing process, the metastable and stable S phase are gradually precipitated from the supersaturated solid solution [10–13], with a sequence: SSSS → GP zone → S′′ → S′ → S → θ. Silcock [14] observed that GP zone was a round particle with a size of only 1–2 nm. The formation of GP zones could significantly improve the mechanical properties of Al–Cu–Mg alloys [16–19]. When ageing at 170 °C for 5 to 10 h, GP zone is one of the main ageing precipitates [20]. Ying et al. [21] reported that the pre-strain working treatment on Al–Cu–Mg alloy significantly increased the size of Cu–Mg atomic clusters during the ageing process, and hence significantly improved the strength. Wang et al. [22] also reported that 2618 Al alloy after strain-ageing treatment had higher room temperature/high temperature strengths, mainly due to the formation of fine S′ particles after plastic deformation. Yu et al. [23] thought that the reason that 2618 alloy had a higher high-temperature performance than other 2-series aluminum alloys was partially due to the formation of the precipitated phase S′. Marquis et al. [24] thought that the improvement in room temperature strength and high temperature creep resistance by Mg addition in Al-Sc alloy was owing to the solid-solution strengthening of Mg solute. Li et al. [25] studied the strengthening mechanisms for age strengthening peak of the CR-processed samples of Al–Cu–Mn alloy with Mg addition are the enhanced particle density and finer S′ distribution besides the strain hardening and grain refinement. Geng et al. [26] studied the effect of Mg in Al–Cu–Mn high-strength foundry alloy at room temperature.0.21% Mg combined with 0.3%Ag are added into Al–6.0%Cu–0.3%Mg alloy to increase the high temperature mechanical properties due to the formation of Ω phase [27]. Xiao et al. [28] studied that Mg added into Al–Cu–Mn–Ag alloy can increase the room temperature and high temperature mechanical properties of the alloy because of the formation of the Ω phase. Researchers have added some elements to Al–Cu–Mn alloy to prompt the formation of intermetallic compounds to increase the high temperature mechanical properties. But the heat-resistant phase may form during solidification, during solution treatment or during ageing treatment, which makes the key contribution to the heat-resistant alloy is not discussed.

So, in this study, Mg is added into Al–Cu–Mn–Ni alloy to affect three types of heat-resistant phases (formed during solidification, solution treatment and ageing treatment respectively), and their contributions to high temperature tensile and yield strength were discussed in detail. New points of views on microstructure design are obtained, which may be in favor of developing new heat-resistant aluminum alloys.

### 2. Materials and method

| Alloy      | Cu (wt%) | Mn (wt%) | Mg (wt%) | Fe (wt%) | Ni (wt%) | Al (wt%) |
|------------|----------|----------|----------|----------|----------|----------|
| A2-1       | 4.70     | 0.99     | 0.001    | 0.093    | 0.490    | Bal.     |
| A2-1-0     | 4.74     | 1.04     | 0.215    | 0.098    | 0.482    | Bal.     |
| A2-1-1     | 4.72     | 1.02     | 0.514    | 0.097    | 0.485    | Bal.     |
| A2-1-2     | 4.71     | 1.03     | 0.795    | 0.097    | 0.487    | Bal.     |

Nominal Al-5wt%Cu-1wt%Mn-0.5 wt%Ni (A2-1), Al-5wt%Cu-1wt%Mn-0.5 wt%Ni-0.2 wt%Mg(A2-1-0), Al-5wt%Cu-1wt%Mn-0.5 wt%Ni-0.5 wt%Mg(A2-1-1) and Al-5wt%Cu-1wt%Mn-0.5 wt%Ni-0.5 wt%Mg (A2-1-2) alloys were prepared in a melting/vacuum holding furnace. The raw materials were from pure Al and Mg ingots and Al-20 wt% Cu, Al-10 wt% Mn and Al-10 wt% Ni master alloy ingots. MAX-LMF15 spectrum was used to measure the chemical composition of the prepared alloys, as listed in Table 1. After melt processing, the melts at 720 °C were finally poured into a cast iron mold with a plate-like cavity size of 170 mm × 100 mm × 20 mm (pre-heated at 250 °C for 5 h). Optical microscope (OLYMPUS BX-60 M), Sirion field-emission scanning electron microscope (SEM) with GENESIS 60 S x-ray EDS and transmission electron microscopy (Tecnai-G20) with the same model EDS were used to characterize the microstructure evolution in samples of
solution and ageing treatments play a very important role in the high temperature strengthening. The alloying of 20 MPa, while that from the ageing treatment is 26 MPa. Therefore, the microstructure evolutions during the solution treatment is 29 MPa and the increment from the ageing treatment is a startling rise of 42 MPa. It increased. For A2-1 alloy without Mg addition, the increment in YS at 300°C increased. For A2-1 alloy at as-solutionized state, alloying of 0.5 wt% Mg results in a great rise in YS at 300°C from 91 MPa to 121 MPa by about 33%, and at as-aged state from 117 MPa to 163 MPa by about 40%. This indicates alloying of Mg exhibits a very strong high temperature strengthening effect on Al which significantly increases yield strength.

Table 2. Data of mechanical properties of A2-1-0, A2-1-1, A2-1-2 alloys in T6 temper, tensile tested at 300°C.

| Alloys  |  A2-1-0 | A2-1-1 | A2-1-2 |
|---------|---------|--------|--------|
| T6      | 156 ± 10| 170 ± 7.0 | 155 ± 6 |
| $\sigma_{0.2}$ | 145 ± 10 | 163 ± 2.8 | 143 ± 4.8 |
| $\delta$ | 4.7 ± 0.8 | 2.4 ± 0.4 | 3 ± 0.7 |

Table 3. Tensile mechanical properties at 300°C of A2-1 and A2-1-1 alloys at different states.

| Alloy   | State     | UTS (MPa) | YS (MPa) | EL (%) | Increment in YS by solution treatment (MPa) | Increment in YS by ageing treatment (MPa) |
|---------|-----------|-----------|----------|--------|--------------------------------------------|-------------------------------------------|
| A2-1    | As-cast   | 96 ± 6    | 71 ± 2   | 20.2 ± 1.7 | 20                                        | 26                                        |
|         | As-solutionized | 119 ± 1.0 | 91 ± 4   | 9.1 ± 1.7 |                                            |                                           |
|         | As-aged   | 141 ± 5.0 | 117 ± 5.0 | 4.7 ± 1.0 |                                            |                                           |
| A2-1-1  | As-cast   | 111 ± 1.0 | 92 ± 2.2 | 4.45 ± 1.1 | 29                                        | 42                                        |
|         | As-solutionized | 125 ± 6.4 | 121 ± 2.8 | 1.4 ± 0.3 |                                            |                                           |
|         | As-aged   | 170 ± 7.0 | 163 ± 2.8 | 2.4 ± 0.4 |                                            |                                           |

A2-1 and A2-1-1 alloys. The solution treatment was performed at 525°C for 6 h followed by quenching in water (25°C) (as-solutionized state), and then the ageing treatment was performed at 170°C for 4 h (as-aged state). According to GB/T228-2002 (a Chinese standard), tensile test samples with a gauge size of 18 × 3 × 3 mm³ were prepared from the castings without or after heat treatments. The tensile test was carried out at 300°C on a CMT4503 electronic universal testing machine with a rate of 1 mm min⁻¹. The average value of 3 testing samples was used to obtain the ultimate tensile strength (UTS) and yield strength (YS). Before the tensile test at 300°C, the samples were first preheated at 300°C for 30 min in test chamber, and then the tensile test was proceeded with a duration time of about 5 min. A sample of A2-1 alloy at as-aged state was selected to perform a heat exposure test at 300°C for 30 min to evaluate the microstructure evolution during high temperature tensile test.

3. Results and discussion

3.1. Reason for choosing the A2-1 alloy with 0.5 wt% manganese addition

0.215% magnesium (A2-1-0 alloy), 0.514% magnesium (A2-1-1 alloy) and 0.795% magnesium (A2-1-2 alloy) are added into Al–5Cu–1Mn–0.5Ni (A2-1) alloy. It is found that the alloy with 0.514% magnesium (A2-1-1 alloy) has the best tensile strength and yield strength at 300°C in T6 temper. So A2-1-1 alloy is chosen as the main studied alloy. Their compositions are listed in Table 1 and the mechanical properties at 300°C are listed in Table 2.

3.2. Mechanical properties at 300°C of A2-1 and A2-1-1 alloys at different states

Table 3 summarizes the tensile mechanical property data of A2-1 and A2-1-1 alloys tested at 300°C at as-cast, as-solutionized and as-aged states. Representative engineering stress-engineering strain curves of them are illustrated in figure 1. There are great differences in UTS, YS and EL (fracture elongation) between these alloys. Obviously, it is due to the alloying effect of Mg.

For as-cast samples, the addition of 0.5 wt%Mg in Al–5Cu–1Mn–0.5Ni alloy makes YS at 300°C rise from 71 MPa to 92 MPa by about 30%. At as-solutionized state, alloying of 0.5 wt% Mg results in a large rise in YS at 300°C from 91 MPa to 121 MPa by about 33%, and at as-aged state from 117 MPa to 163 MPa by about 40%. This indicates alloying of Mg exhibits a very strong high temperature strengthening effect on Al–Cu–Mn–Ni alloy at three states. It should be noted that alloying of Mg also leads to a considerable decrease in fracture elongation at high temperature simultaneously.

From Table 3, it is also seen that both solution treatment and ageing treatment make YS at 300°C greatly increased. For A2-1 alloy without Mg addition, the increment in YS at 300°C from the solution treatment is 20 MPa, while that from the ageing treatment is 26 MPa. Therefore, the microstructure evolutions during the solution and ageing treatments play a very important role in the high temperature strengthening. The alloying of Mg further enhances this effect. For A2-1-1 alloy with 0.5 wt%Mg addition, the increments in YS at 300°C from the solution treatment is 29 MPa and the increment from the ageing treatment is a startling rise of 42 MPa. It
implies that alloying of Mg has a remarkable influence on the microstructure evolution during the solution and ageing treatment processes, especially the latter.

3.3. As-cast microstructure by alloying of Mg
In the alloy without Mg addition (A2-1), the eutectic structure is composed of skeleton and blocky Ni-rich phase, flocculent \( \theta \)-CuAl\(_2\) phase and relatively coarse blocky \( T_{\text{Mg}(\text{Al}_{20}\text{Cu}_{3}\text{Mn}_{1})} \) phase, distributed along eutectic boundaries \[8\]. Figure 2(a) shows the eutectic microstructure in the A2-1-1 alloy with 0.5 wt% Mg addition. Alloying of Mg results in considerable changes of the finally solidified eutectic structure. On one hand, the eutectic \( \theta \)-Al\(_2\)Cu phase changes from un-compacted flocculent to compacted blocky, as seen in figure 2(b). On other hand, there are a great amount of Mg-rich phase (Al–Cu–Mg compound) observed, with two

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**Figure 1.** Representative engineering stress—engineering strain curves of Al–5Cu–1Mn–0.5Ni alloy with/without alloying of Mg at different states: (a) as-cast; (b) as-solutionized; (c) as-aged.

**Figure 2.** Optical microstructure of Al–5Cu–1Mn–0.5Ni–0.5Mg alloy (A2-1-1) as-cast: (a) low magnification; (b) high magnification.
morphologies: spherical particulates and whiskers in ridge shape (seen in figure 2(b)). The spherical particle, in fact, has an un-compacted multi-phase structure (figure 3(a), marked by Arrow 1). Its EDS results (figure 3(b)) contain high level of Mg, indicating that it may be composed of Al-Cu-Mg, CuAl2, and Al phases. EDS results in figure 3(c) suggest that the compacted blocky phase (marked by Arrow 2 in figure 3(a)) is Al2Cu phase. TEM observation captures the black whisker phase in figure 2(b). Its morphology is exhibited in figure 4(a), presenting as lamelliform. The EDS results (figure 4(b)) and SAED pattern (figure 4(c)) confirm that this whisker-shape phase is AlCuMg intermetallic compound that has a hexagonal structure with a = 0.6 nm, b = 0.6 nm, c = 0.71 nm. However, S-(Al12CuMg) compound, usually found in Al-Cu-Mg alloys, has an orthorhombic structure with a = 0.4 nm, b = 0.925 nm, c = 0.715 nm [29, 30]. In the Al–5Cu–1Mn–0.5Ni–0.5Mg alloy studied, the formed Mg-rich phase is AlCuMg compound, not S-(Al12CuMg). The whisker-shape AlCuMg phase abundantly exists at the eutectic boundary zones, which can hinder the boundary gliding and grain rotating during deformation at high temperature.

In summary, alloying of Mg changes the morphology of the eutectic CuAl2 phase from un-compacted to compacted, and leads to the formation of a great amount of whisker-shape AlCuMg phase in the form of ridge-like around the eutectic boundary zone. Thus, it results in a great rise in YS of as-cast samples at 300 °C by nearly 30%.

3.4. As-solutionized microstructure by alloying of Mg
Figure 5 shows the optical microstructure of the as-solutionized Al–5Cu–1Mn–0.5Ni–0.5Mg alloy (A2-1-1). Similar with the A2-1 alloy without the addition of Mg [8], eutectic T_{Mg} phase (black blocky-like particles in figure 5) still exists at the eutectic boundaries as well as Ni-rich phase (dark grey in figure 5). SEM image and EDS results (shown in figure 6) illustrate these Ni-rich phases have been granulated due to partial dissolution. It is always thought that the eutectic boundaries are vulnerable zones at high temperature [31]. The existence of Al2CuNi, Al2Cu3Ni and T_{Mg}(Al120Cu2Mn3) phases with good thermal stability can impede the boundary gliding and grain rotating and thus improve the high temperature strength. At the same time, the eutectic AlCuMg
Figure 4. TEM image (a) showing the morphology of the black whisker-shape phase in figure 2(b) and its EDS results (b) and SAED pattern (c) in as-cast Al–5Cu–1Mn–0.5Ni–0.5Mg alloy.

Figure 5. Optical microstructure of as-solutionized Al–5Cu–1Mn–0.5Ni–0.5Mg alloy (A2-1-1).

Figure 6. SEM image (a) of as-solutionized Al–5Cu–1Mn–0.5Ni–0.5Mg alloy (A2-1-1) and corresponding EDS results of Ni-rich phases (b).
whiskers and spherical particles existing at the eutectic zones have completely dissolved into the aluminum matrix as well as the compacted eutectic CuAl2 phase. As a result, there are a great area of band-like blank zones left (free precipitation zones) (seen in figure 5).

Figure 5 also exhibits a great many of dispersoid particles in Al matrix precipitated during the solution treatment of the alloy with Mg addition, which has no obvious difference with the alloy without Mg [8]. TEM image in figure 7(a) shows the morphology of these precipitated particles. EDS results in figures 7(b)–(c) illustrate that they are Al–Cu–Mn compounds with two atomic ratios of 2:3 and 3:2, Al15Cu3Mn2 (usually named by TMn′) and AlCu3Mn2 respectively, which have been determined by SAD patterns in [8]. From TEM observation, alloying of Mg has little influence on the precipitation behavior of these Al–Cu–Mn compounds. These precipitates existing in the matrix can impede the dislocation motion during the tensile deformation at high temperature and thus strengthen the matrix.

During the solution treatment, CuAl2 and AlCuMg phases are completely dissolved into the matrix. Partial Cu solute atoms are consumed to form TMn′ and AlCu3Mn2 dispersoid particles. Almost all Mg atoms and residual Cu atoms have entered aluminum solution as solutes. The radius of Mg atom (0.16 nm) is larger than Al (0.143 nm). Thus, Mg entering Al solution will lead to lattice distortion, which enlarges the difficulty of the motion of the dislocations. Of more importance, solute atoms of Mg and Cu as well as the lattice distortion induced by them act with the dislocations that form Cottrell atmosphere. This strongly pins the dislocations, hindering the dislocation motion.

Microstructure observation indicates that alloying of Mg has little influence on the precipitation behavior during the solutionizing process, however, the complete dissolution of AlCuMg compound which is formed during solidification raises the solubility of Mg in the matrix. The solution strengthening from Mg results in a considerable rise (by about 30 MPa) in YS at 300 °C of the alloy with Mg addition (in T4 temper), compared with the alloy without Mg addition. Simultaneously, the increment in YS from the solution treatment for the alloy with the addition of Mg is also obviously increased to 29 MPa from 20 MPa (without Mg addition).

3.5. Effect of alloying of Mg on microstructure at as-aged state and after tensile test at 300 °C
After the ageing treatment at 170 °C, high density of fine \(\theta''\) particles are precipitated in the Al matrix of A2-1 alloy (without Mg addition) [8] that lead a considerable increment in YS at 300 °C from the ageing treatment (26 MPa). In A2–1–1 alloy (with Mg addition), there are also high density of \(\theta''\) particles precipitated and their size become finer (figure 8(a)). Surprisingly, a great number of annular ring structures are also observed, as seen in figure 8(b). This type of structure was not observed in the studied Al–5Cu–1Mn and Al–5Cu–1Mn–0.5Ni alloys [8]. The high magnification image (figure 8(c)) indicates that these rings seem to be composed of the pinned dislocations, and there are some suspected needle-like and cube-like particles inside or outside the rings. SAED pattern shown in figure 8(d) doesn’t reveal the existence of any precipitates at these ring zones. Ringer et al [32] also observed similar annular rings in Al–1.1Cu–1.7Mg–0.1Ag alloy at as-aged state and thought that these...
were typical Cu-Mg atomic clusters by APT analysis. After a long time holding at the solution temperature, a great number of vacancies are retained in the aluminum solution during fast cooling by quenching into water. That is, the vacancy is supersaturation in the matrix after the solution treatment. At the same time, the solutes of Mg and Cu atoms are supersaturated too. The interaction of vacancies and solute atoms leads to the formation of atomic clusters\(^3\).

Figure 9\(^{(a)}\) show TEM microstructure observation near the fractured surface of the sample of A2-1-1 alloy (at as-aged state) after the tensile test at 300\(^{\circ}\)C. Because the sample had undergone a 30 min holding at 300\(^{\circ}\)C before the tensile test, \(\theta''\) phase formed during the ageing course has transformed into \(\theta'\) phase (marked by Arrow 1 in figure 9\(^{(a)}\)) which is determined by SAED pattern in figure 9\(^{(b)}\). And the size is finer than that in the sample of A2-1 alloy (seen in figure 10) after an exposure at 300\(^{\circ}\)C for 30 min. Alloying of Mg seems to improve the thermal stability of \(\theta'\), because it is not greatly coarsened like in A2-1 alloy. Of more importance, in the sample of A2-1-1 alloy after the tensile test at 300\(^{\circ}\)C, there are numerous of cube-like particles formed (marked by Arrow 2 in figure 9\(^{(a)}\)). They are about 40 nm in size. Its EDS results (figure 9\(^{(c)}\)) indicate that it is composed of Al, Cu and Mg. According to the calculation of SAED pattern (figure 9\(^{(d)}\)), this cube particle is determined as \(\text{Al}_5\text{Cu}_2\text{Mg}\), with an orthorhombic structure of \(a = 0.61\) nm, \(b = 0.79\) nm, \(c = 1.84\) nm, paralleling with \([020]\), \([131]\) and \([111]\) of Al matrix. Obviously, this cube particle is different from S-\(\text{Al}_5\text{CuMg}\) phase which has an orthorhombic structure with \(a = 0.4\) nm, \(b = 0.925\) nm, \(c = 0.715\) nm [29, 30].

Compared with A2-1 alloy, there are a great number of Cu-Mg clusters formed in A2-1-1 alloy (with Mg addition) after the ageing treatment besides a large amount of fine precipitated \(\theta''\) phase. And after 30 min’s holding at 300\(^{\circ}\)C before the tensile test, the size of \(\theta'\) phase transformed from \(\theta''\) is finer. Especially, there are a great number of fine cube-like particles formed. These precipitates severely impede the sliding of dislocations on the lattice planes and thus strengthen the matrix greatly. Therefore, the yield strength at 300\(^{\circ}\)C of A2-1-1 alloy at as-aged state is much higher than that of A2-1 alloy by 46 MPa. It is the significant microstructural evolution induced by alloying of Mg that leads to the great rise in the increment of yield strength at 300\(^{\circ}\)C from the ageing treatment, high to 42 MPa. This again demonstrates the important role of the finer particles precipitated during the ageing treatment during the high temperature strengthening.
4. Conclusions

(1) The alloying with 0.5 wt%Mg addition in Al-5Cu-1Mn-0.5Ni alloy considerably increases the tensile strength and yield strength at 300 °C of the samples at as-cast, as-solutionized and as-aged states. And it greatly enhances the strengthening effects from the solution and ageing treatments, however, simultaneously decreases the elongation greatly.

Figure 9. TEM microstructure observation near the fractured surface of A2-1-1 alloy (at as-aged state) after the tensile test at 300 °C (a), SAED pattern of the precipitated phase marked by Arrow 1 (b), EDS results (c) and SAED pattern (d) of the cube particle marked by Arrow 2.

Figure 10. TEM image showing the coarse precipitates in A2-1 alloy (at as-aged state) after 300 °C exposure for 30 min (a) and its SAED pattern.
(2) During the solidification of the alloy with the alloying of Mg, a great amount of whisker-like Al–Cu–Mg compound phase is formed at the eutectic boundary zones in the form of ridge configuration. By TEM characterization, this lamelliform compound is determined as AlCuMg with a hexagonal structure of a = 0.6 nm, b = 0.6 nm, c = 0.71 nm.

(3) The alloying of Mg has little influence on the formation of the eutectic T_{Al} and Ni-rich phases. But it leads to the formation of whisker-like AlCuMg phase in the form of ridge configuration at eutectic boundary zones and makes the eutectic CuAl2 phase more compacted. Therefore, it greatly increases the strength of the alloy at as-cast state at 300 °C.

(4) The alloying of Mg has little influence on precipitation behaviors of T_{Mn}′ (Al_{20}Cu_{2}Mn_{3}) and AlCu_{3}Mn_{2} phases during the solution treatment. But the whisker-like AlCuMg phase is completely dissolved that increases the solubility of Mg in Al matrix. As a result, it greatly increases the tensile strength and yield strength at 300 °C of the alloy at as-solutionized state and leads to remarkable rise in the increment of yield strength at 300 °C from the solution treatment.

(5) The alloying of Mg has a significant influence on the precipitation behaviors during the ageing treatment, for examples, promoting the formation of finer θ′ phase and a great number of Cu–Mg atomic clusters. During the high temperature tensile test, theses clusters will evolve into fine cube-like Al_{3}Cu_{2}Mg particles with an orthorhombic structure, and the size of θ′ phase transformed from θ′′ is finer too. It is the significant microstructure evolution induced by the alloying of Mg that leads to the great rise in the increment of tensile strength and yield strength at 300 °C from the ageing treatment, high to 45 MPa and 42 MPa respectively. This again demonstrates the important role of the finer particles precipitated during the ageing treatment in the high temperature strengthening.

(6) The addition of Mg in Al-Cu-Mn-Ni alloy results in the formation of whisker-like AlCuMg compound during solidification and cube-like Al_{3}Cu_{2}Mg particles during the tensile test at 300 °C, rather than metastable or stable S phase.

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**Data availability statement**

All data that support the findings of this study are included within the article (and any supplementary files).

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**References**

[1] Sha G, Marceau R K W, Gao X, Muddle B C and Ringer S P 2011 Nanostructure of aluminum alloy 2024: segregation, clustering and precipitation processed Acta Mater. 59 1659–70
[2] Mohamed A M A, Samuel F H and Kahtani S A 2013 Microstructure, tensile properties and fracture behavior of high temperature Al–Si–Mg–Cu cast alloys Mater. Sci. Eng. A 577 64–72
[3] Miao W F and Laughlin D E 2000 Effects of Cu content and presaging on precipitation characteristics in aluminium alloy 6022 Metall. Mater. Trans. A 31 361–71
[4] Zhao B B, Zhan Y Z and Tang H Q 2019 High-temperature properties and microstructural evolution of Al–Cu–Mn–RE(La/Ce) alloy designed through thermodynamic calculation Mater. Sci. Eng. A 758 7–18
[5] Zhang X M, Wang W T, Chen M A, Gao Z G, Jia Y Z, Ye L Y, Zheng D W, Liu L and Kuang X Y 2010 Effects of Yb addition on microstructures and mechanical properties of 2519A aluminum alloy plate Trans. Nonferro. Met. Soc. Chin. 20 727–31
[6] Wang W T, Zhang X M, Gao Z G, Jia Y Z, Ye L Y, Zheng D W and Liu L 2010 Influences of Ce addition on the microstructures and mechanical properties of 2519A aluminum alloy plate J. Alloy Compd. 491 366–71
[7] Lin B, Zhang W X, Zheng X P, Zhao Y L, Lou Z H and Zhang W W 2019 Developing high performance mechanical properties at elevated temperature in squeeze cast Al–Cu–Mn–Fe–Ni alloys Mater. Charact. 150 128–37
[8] Chen J L, Liao H C, Wu Y N and Li H L 2020 Contributions to high temperature strengthening from three types of heat-resistant phases formed during solidification, solution treatment and aging treatment of Al–Cu–Mn–Ni alloys respectively Mater. Sci. Eng. A 772 138819
[9] Roósz A and Exner H E 1990 Ternary restricted-equilibrium phase diagrams: II. Practical application: aluminum-rich corner of the Al–Cu–Mg system Acta Metall. Mater. 38 2009–16
[10] Williams J C and Starke E A 2003 Progress in structural materials for aerospace systems Acta Mater. 51 5775–99
[11] Khan I N, Starink M J and Yan J L 2008 A model for precipitation kinetics and strengthening in Al–Cu–Mg alloys Mater. Sci. Eng. A 472 66–74

[12] Wang S C, Starink M J and Gao N 2006 Precipitation hardening in Al–Cu–Mg alloys revisited Scripta Mater. 54 287–91
[13] Parel T S, Wang S C and Starink M J 2010 Hardening of an Al–Cu–Mg alloy containing Types I and II S phase precipitates Mater. Des. 31 S2–5

[14] Silcock J M 1961 The structural ageing characteristics of Al–Cu–Mg alloys with copper-magnesium weight ratios of 7-1 and 2.2–1 J. Inst. Met. 89 203–10 [https://webofscience.com-s.vpn.seu.edu.cn:8118/wos/alldb/full-record/WOS:000088372100014]
[15] Hardy H K and Heal T J 1954 Report on precipitation Prg Metal Phys. 5 143–278
[16] Ringer S P, Hono K, Sakurai T and Polmear I J 1997 Cluster hardening in an aged Al–Cu–Mg alloy Scripta Mater. 36 517–21

[17] Ringer S P, Sakurai T and Polmear I J 1997 Origins of hardening in aged Al–Cu–Mg–(Ag) alloys Acta Mater. 45 3731–44

[18] Marceau R K W, Sha G, Ferragut R, Dupasquier A and Ringer S P 2010 Solute clustering in Al–Cu–Mg alloys during the early stages of elevated temperature ageing Acta Mater. 58 4923–39

[19] Deschamps A A, Bastow T J, de Geuser F, Hill A J and Hutchinson C R 2011 In situ evaluation of the microstructure evolution during rapid hardening of an Al–2.5Cu–1.5Mg (wt%) alloy Acta Mater. 59 2918–27

[20] Liu M, Bai S, Liu Z Y, Zhou X W, Xia P and Zeng S M 2015 Analysis of modulus hardening in an artificially aged Al–Cu–Mg–Ag alloy by atom probe tomography Mater. Sci. Eng. A 629 23–8

[21] Ying P Y, Liu Z Y, Bai S, Liu M, Lin L H, Xia P and Xia L Y 2017 Effects of pre-strain on Cu–Mg co-clustering and mechanical behavior in a naturally aged Al–Cu–Mg alloy Mater. Sci. Eng. A 704 18–24

[22] Wang J H, Yi D Q, Su X P and Yin F C 2008 Influence of deformation ageing treatment on microstructure and properties of aluminum alloy 2618 Mater. Charact. 59 965–8

[23] Yu K, Li S R and Li W X 2000 Effect of Trace Sc and Zr on the mechanical properties and microstructure of Al–alloy 2618 J. Mater. Sci. Technol. 16 416–20 [https://webofscience.com-s.vpn.seu.edu.cn:8118/wos/alldb/full-record/WOS:000088372100014]

[24] Marquis E A, Seidman D N and Dunand D C 2003 Effect of Mg addition on the creep and yield behavior of an Al–Sc alloy Acta Mater. 51 4751–60

[25] Li H Z, Liu R M, Liang X P, Deng M, Liao H J and Huang I L 2016 Effect of pre-deformation on microstructures and mechanical properties of high purity Al–Cu–Mg alloy Trans. Nonferrous Met. Soc. China 26 1482–90

[26] Geng H N, Zhang B R and Ma J J 1994 Effects and behavior of magnesium in Al–Cu–Mn high-strength foundry alloy Foundry 9 15–19, 34 [https://kdoc.cnki.net/s.vpn.seu.edu.cn:8118/kdoc/docdown/pubdownload.aspx?dk=kdoc%3apdfdown%3aa52ef73fb0c060d4f0dc80004a718&lang=G8]

[27] Li H Z, Zhang X M, Chen M A and Zhou Z P 2006 Effects of Ag on microstructure and mechanical properties of 2519 aluminum alloy J. Cent. South Univ. Technol. 13 130–4

[28] Xiao D H, Huang B Y and Chen K H 2007 Effect of Mg addition on microstructure and mechanical properties of Al–Cu–Mn–Ag alloy Mater. & Heat Treatment 36 1–3, 6

[29] Charai A, Walther T, Alfonso C, Zahra A M and Zahra C Y 2000 Coexistence of clusters, GPB zones, S’-, S’- and S-phases in an Al–0.9% Cu–1.4% Mg alloy Acta Mater. 48 2751–64

[30] Kovarik L, Court S A, Fraser H L and Mills M J 2008 GPB zones and composite GPB/GPBI zones in Al–Cu–Mg alloys Acta Mater. 56 4804–15

[31] Qian Z Y, Liu X F, Zhao D G and Zhang G H 2008 Effects of trace Mn addition on the elevated temperature tensile strength and microstructure of a low–iron Al–Si piston alloy Mater. Lett. 62 2146–9

[32] Ringer S P, Hono K, Polmear I J and Sakurai T 1996 Nucleation of precipitates in aged Al–Cu–Mg–(Ag) alloys with high Cu:Mg ratios Acta Mater. 44 1883–98

[33] Ivanov R, Deschamps A and De F 2018 Geuser, Clustering kinetics during natural ageing of Al–Cu based alloys with (Mg, Li) additions Acta Mater. 157 186–95