Uneven Precipitation Behavior during the Solutionizing Course of Al-Cu-Mn Alloys and Their Contribution to High Temperature Strength

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The dispersoid precipitation behavior during the solutionizing and aging of Al-xwt.%Cu-1.0 wt.% Mn alloys (x = 2.0, 4.5, and 7.5) and contribution to mechanical properties were investigated using tensile testing and microstructural characterization. A shell-core structure of primary α-Al dendrites is found in Al-Cu-Mn alloys, in which the Cu content in the shell is higher than that in the core. The area of shell zone (Cu-rich) increases with an increase in Cu content in the alloy. Large amounts of fine dispersoid Al-Cu-Mn particles precipitate in solution. An alloy with low Cu content results in only the $\tilde{T}$Mn($\text{Al}_{20}\text{Cu}_{2}\text{Mn}_3$) particles being precipitated. However, in an alloy with high Cu content, $\text{AlCu}_3\text{Mn}_2$ particles are first found to precipitate beside $\tilde{T}$Mn. However, this precipitation behavior is uneven. The precipitation zones in the solution microstructure are consistent with the Cu-rich regions in the as-cast microstructure. A number of fine particles (dozens nanometer in size) are first found to precipitate on the rod-like $\tilde{T}$Mn particles during the aging phase. The redissolution and granulation of the eutectic CuAl$_2$ phase during the solutionizing process result in the formation of particle-free bands between the precipitation zones. The tension test at 300°C demonstrates that the increase in high temperature strength is due to the dispersoid precipitation during solutionizing, and the precipitation behavior in the aging phase has little or no effect, however, largely improves the tensile strength at room temperature. High temperature strength is significantly increased with an increase in Cu content, which correlates to an increase in number and decrease in size of $\tilde{T}$Mn and $\text{AlCu}_3\text{Mn}_2$ particles.

1. Introduction

Because of their good high temperature performance, Al-Cu-Mn alloys, such as 2519 and 2219 as well as A2219, have been used in structural parts for auto and space vehicles working at higher temperature environment [1–4]. In order to further improve the high-temperature mechanical properties, controlling the size and number of strengthening precipitates with high thermal stability is of great importance.

In Al-Cu-Mn alloys, the main strengthening phases at high temperature have been proposed to be the $\tilde{T}$Mn and $\theta$ phases. For 2 × 24 alloys, the interaction between the dislocation motion and dispersoid precipitates of $\tilde{T}$Mn($\text{Al}_{20}\text{Cu}_{2}\text{Mn}_3$) and $\theta$ (CuAl$_2$) phases play key roles in strengthening at high temperature [5–8]. For the 2 × 24 alloys (Al-Cu-Mg-Mn alloys), large amounts of rod-like T-phase ($\text{Al}_{19}\text{Cu}_{12}\text{Mn}_3$) dispersoid particles are precipitated in ingot homogenization or solution treatment processes [9, 10]. Chen et al. [11] reported that the overwhelming majority of particles of the $\tilde{T}$Mn phase take a shape of a rod in the Al-5Cu-1Mn alloy after solutionizing. Chen et al. [12] produced Al-4.6Cu-Mn ribbons using melt spinning, and during this process, rod-like $\tilde{T}$Mn dispersoids with a diameter of 120–160 nm formed during the aging process at 190°C. The work of Wang et al. [13] was in agreement, reporting a rod-like $\tilde{T}$Mn phase in Al-Cu-Mn alloys that have the lattice parameters $a = 2.41$ nm, $b = 1.25$ nm, $c = 0.78$ nm. Using CBED (convergent beam electron diffraction), Park and Kim [14] determined the structure parameters of the $\tilde{T}$Mn phase of $a = 2.345$ nm,
proposed that the $T_{\text{Mn}}$ phase belongs to a space group of BBMM (Brownian bridge movement model) with lattice parameters of $a = 23.98 \text{ Å}$, $b = 12.54 \text{ Å}$, and $c = 7.66 \text{ Å}$ using near-atomic resolution chemical mapping. Feng et al. [9] adopted HRTEM and high-angle annular dark-field scanning transmission electron microscopy (HAADF-STEM) to confirm that the $T_{\text{Mn}}$ phase has an orthorhombic structure. Many studies report that the dispersed Al-Cu-Mn compound particles precipitated during the solutionizing are the $T_{\text{Mn}}$ phase, but another crystal structure of an Al-Cu-Mn compound with formula AlCu$_3$Mn$_2$ was proposed as a single-phase system using X-ray and neutron diffractometer traces [15]. There are many reports in the literature focusing on the crystal structure of Al-Cu-Mn precipitates. However, the conditions during precipitation from a supersaturated aluminum solution and strengthening effect are not understood and require further study.

PFZ (precipitation-free zone) often occurs at the grain boundaries in Al-Cu alloys. It is widely accepted that it is the escaping of both the vacancies and solute atoms of Cu into the grain boundary during aging course that leads to the formation of PFZ. The formation of PFZs has an important influence on mechanical properties, especially on antistress corrosion property. Li et al. [16] reported that the width of the PFZs became larger with increasing aging time of the Al-Zn-Mg-Sc-Zr alloy. Muntiz and Cotler [17] thought that the smaller the width of PFZs is, the higher the tensile strength is. Both the precipitation behaviors of $T_{\text{Mn}}$ particles during solutionizing and $\theta$ particles during aging are formed from Al solution; however, it is not clear whether PFZs occur during solutionizing.

In this study, the microstructure and mechanical properties of three Al-xwt.%Cu-1.0 wt.% Mn alloys at ambient and elevated temperatures were investigated to find the precipitation behavior of $T_{\text{Mn}}$ and AlCu$_3$Mn$_2$ phases in the Al-Cu-Mn system alloy and their contribution to the mechanical properties of the alloy.

2. Experimental Procedures

Three nominal Al-xwt.%Cu-1.0Mn alloys (x = 2.0, 4.5, and 7.5) were prepared by melting pure Al (99.7%) ingot and Al-20Cu and Al-10Mn master alloys in a resistance-heated furnace with the graphite crucible of 3 kg capacity. The prepared alloys are marked as A$_1$, A$_2$, and A$_3$, respectively, for 2, 4.5, and 7.5 wt. % of Cu contents. MAX LMF15 spectrum was used to measure the chemical composition of the studied alloys, as listed in Table 1. After processing, the melts were poured into a metal mold with a cavity size of 170 mm x 100 mm x 20 mm which was preheated at 250°C for at least 5 h.

Parts of the obtained castings underwent a heat treatment: solutionized at 525°C for 6h + 535°C for 6h (a modified T6 state), followed by quenching in warm water and then aging at 170°C for 4 hrs. Metallography samples were cut from the castings. Optical microscope, SEM with GENESIS 60S X-ray EDS, and transmission electron microscopy (TEM) were applied to characterize the microstructure of as-cast, as-solutionized, and T6 samples. Tensile test samples (based on Chinese standard: GB/T228-2002) were also cut from the castings with a gauge size of 35 x 10 x 3 mm$^3$. The tensile testing was performed on a CMT4503 electronic universal testing machine with a rate of 1 mm/min at 25°C and 300°C. The data of mechanical properties are taken as an average of 3 samples.

3. Results and Discussion

3.1. Precipitation Behaviors during Solutionizing. Figure 1 shows the microstructure of the studied alloys as-solutionized. At the first sight, it seems that these optical photos are not qualified. Actually, it is a great number of precipitates that lead to this fuzziness. High-magnification photo (Figure 1(d)) clearly exhibits the fine dispersed particles precipitated during the solutionizing course of the Al-Cu-Mn alloys. Here, the area where precipitation particles exist is designated as the precipitation zone (PZ). Of more importance, this precipitation behavior is uneven. The contour of the primary dendrites of the $\alpha$-Al phase is dimly discernible, meaning that no or few particles are precipitated in the inner of dendrites (here, named as no particle zones, NPZs). It is seen that the area of NPZs is decreased with Cu content in the alloy; that is, the area of PZs is increased. And also there are a number of continuous or semicontinuous PFBs (precipitation-free bands) among the PZs. SEM image of the A$_1$ alloy as-solutionized more powerfully illustrates this precipitation characteristic (Figure 2(a)). The EDS results (Figure 2(b)) suggest these precipitated particles contain high level of Al, Cu, and Mn, being an Al-Cu-Mn compound. Figure 2 also shows a PFB (marked by Arrow 3) and CuAl$_2$ particle that are survived from the solutionizing course.

Figure 3 shows the TEM images and EDS results of the precipitates in the studied alloys as-solutionized and SAD patterns of representative precipitation particles. TEM images exhibit that most of them take a shape of the rod, long or short in size, consistent with [9–11, 18]. Usually, the composition detected by TEM-EDS is much more accurate than that by SEM-EDS. In the A$_1$ alloy with a low Cu content of about 2 wt. %, TEM-EDS result indicates the precipitation particles are Al-Cu-Mn compound with a Cu/Mn atomic ratio close to 2:3 (Figure 3(b)). Feng et al. [9], Shen et al. [10], and Chen et al. [11] thought the rod-like particles precipitated in the solutionized 2 x 24 alloy and Al-5Cu-1Mn alloy are $T_{\text{Mn}}$ phase with a formula of Al$_{20}$Cu$_2$Mn$_3$. Liao et al. [18] also observed short rod-like dispersed particles of the Al-Cu-Mn compound in the solutionized Al-Si-Cu-Mn alloy and identified it as $T_{\text{Mn}}$ (Al$_{20}$Cu$_2$Mn$_3$) by calculating the SAD patterns. According to the phase diagram of the Al-Cu-Mn system [19],

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Cu</th>
<th>Mn</th>
<th>Si</th>
<th>Mg</th>
<th>Fe</th>
<th>Al</th>
</tr>
</thead>
<tbody>
<tr>
<td>A$_1$</td>
<td>2.04</td>
<td>0.988</td>
<td>0.064</td>
<td>0.0009</td>
<td>0.130</td>
<td>Bal.</td>
</tr>
<tr>
<td>A$_2$</td>
<td>4.54</td>
<td>0.917</td>
<td>0.059</td>
<td>0.0004</td>
<td>0.131</td>
<td>Bal.</td>
</tr>
<tr>
<td>A$_3$</td>
<td>7.41</td>
<td>1.01</td>
<td>0.061</td>
<td>0.0006</td>
<td>0.138</td>
<td>Bal.</td>
</tr>
</tbody>
</table>
in the Cu-rich corner, the most possible phase of Al-Cu-Mn compounds is $T_{\text{Mn}}$. Thus, the precipitates in the A$_1$ alloy are thought to be the $T_{\text{Mn}}$ phase. Calculation of the SAD pattern in Figure 3(c) further demonstrates it, having an orthorhombic structure with lattice parameters $a = 2.420$ nm, $b = 1.250$ nm, and $c = 0.772$ nm that is consistent with Wang’s et al. [13] and Park and Kim’s [14] models. But A$_2$ and A$_3$ alloys with high Cu content are about 4.5 wt.% and 7.0 wt.%, respectively, besides $T_{\text{Mn}}$ precipitates (such as Arrow 2 in Figure 3(e) and EDS in Figure 3(f) and Arrow 4 in Figure 3(i) and EDS in Figure 3(j)), and another Al-Cu-Mn precipitation particles are found (such as Arrow 1 in Figure 3(e) and Arrow 3 in Figure 3(i)), in which the EDS results (Figures 3(d) and 3(h)) indicate they have an almost same Cu/Mn atomic ratio of 3:2. In 1968, Johnston and Hall [15] identified the crystal structure of the Cu$_3$Mn$_2$Al compound in its single-phase system by X-ray and neutron diffractometer traces: it is a cubic Laves phase with the 8 manganese atoms ordered at the geometrically larger A sites, while the copper and aluminum atoms are randomized in the 16-fold B sites, and its parameter is 6.9046 Å. In this study, the calculation of the SAD pattern of this compound (Figure 3(g)) suggests that it has a simple cubic structure with $a = 0.6904$ nm, equal to AlCu$_3$Mn$_2$ (PDF card no.: 23-0760). Thus, these Al-Cu-Mn precipitation particles with a Cu/Mn atomic ratio of 3:2 found in the as-solutionized A$_2$ and A$_3$ alloys are identified as AlCu$_3$Mn$_2$. Now, it is concluded that, in the Al-Cu-Mn alloy with a low Cu content, there are only $T_{\text{Mn}}$ (Al$_{20}$Cu$_2$Mn$_3$) dispersoid particles precipitated during the solutionizing course, but in the alloy with a high Cu

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**Figure 1:** Optical microstructure of the studied alloys as-solutionized, showing PZs (precipitation zones), NPZs (no precipitation zones), and PFBs (precipitation-free bands): (a) A$_1$ alloy; (b) A$_2$ alloy; (c) A$_3$ alloy; (d) large magnification photo of the A$_2$ alloy.

**Figure 2:** SEM image showing the precipitated particles (a) and its EDS results marked by Arrow 1 (b) in the A$_2$ alloy as-solutionized.
Figure 3: Continued.
content, AlCu₃Mn₂ particles are demonstrated to precipitate beside Tₘₐₙ particles. From Figure 3, it is also worth to note that the number of precipitation particles is increased with Cu content in the alloys; however, the particle size is decreased simultaneously, which will exert impact to the tension properties.

To deeply understand this precipitation behavior during solutionizing, uneven precipitation of the Tₘₐₙ phase, and formation of PFBs among the PZs, it is necessary to carefully characterize the microstructure as-cast. Figure 4 shows the optical microstructure of the A₁, A₂, and A₃ alloys as-cast. There are particulate-like or continuous/semicontinuous network constitutes in the interdendritic area. They are eutectic resultants during the final stages of nonequilibrium solidification of the Al-Cu-Mn alloy: \( L \rightarrow α-Al + θ \) and \( L \rightarrow α-Al + θ + Tₘₐₙ \) [19–21]. SEM images of three constitute and their EDS results in Figure 5 demonstrate it. With Cu content in the alloy, the amount of network \( θ \) phase is increased in the eutectic resultants (Figure 5).

It is very interesting to note that, in Figure 4, there are two contrasts on primary \( α-Al \) dendrites, like a shell-core structure. The part in the core of dendrites is partly bright, and the shell around the core is partly grey. In SEM image (Figure 6(a)), the core of dendrites presents as dark-grey and the shell as light-grey. EDS results of two regions are illustrated in Figures 6(b) and 6(c), respectively, suggesting that Cu content in the shell is much higher than that in the core; however, Mn content is almost equal. EPMA diagrams in Figure 7 show the distribution of Cu, Fe, and Mn elements in the microstructure of the A₂ alloy as-cast. It demonstrates again that the shell of dendrites is much more Cu-rich than the core. And except for a few regions with the Tₘₐₙ phase, the other regions where CuAl₂ phase exists alone are highly Mn-poor. Now, the shell region of \( α-Al \) dendrites can be denoted as the Cu-rich region. The Al-rich corner of the Al-Cu-Mn phase diagram is rather complicated [19]. The physic process for the formation of this obvious shell-core structure of primary \( α-Al \) dendrites is still unknown.

Of more importance, the area of Cu-rich regions is increased with Cu content in the alloy (Figure 4). Carefully comparing Figure 1 with Figure 4, it is found in surprise that Cu-rich regions in as-cast microstructure are highly consistent with the precipitation zones in as-solutionized microstructure. It is just in the Cu-rich regions that precipitation of Tₘₐₙ and AlCu₃Mn₂ dispersoid particles occurs. Due to nonequilibrium solidification, a shell-core structure of primary \( α-Al \) dendrites is formed. In Cu-rich regions, Cu content is about 5 wt.% (Figure 6(c)), very close to the maximum solubility of Cu solute in Al solution in the Al-Cu binary system. In other words, these Cu-rich regions are supersaturated actually. In the course of solutionizing at 525°C for 6 hrs + 535°C for 6 hrs, part of the eutectic CuAl₂ phase formed during nonequilibrium solidification is forced to redissolve into the Al matrix. In the A₁ alloy with only 2 wt.% Cu content, the eutectic CuAl₂ phase is dissolved completely (Figure 8(a)); however, in the A₂ and A₃ alloys, there are some eutectic CuAl₂ particles which survived (Figures 8(b) and 8(c), respectively). Eutectic CuAl₂ phase is located in the Cu-rich regions (Figure 4). The redissolution of these regions further enhances the degree of supersaturation. Even at 535°C of solutionizing temperature, it is supersaturated. Thus, it is the supersaturation of the Cu solute in the Al solution that drives the precipitation of Tₘₐₙ and AlCu₃Mn₂ to occur in the Cu-rich regions. Because the diffusion coefficient of manganese in the Al solution is very low \((1 \times 10^{-12}\text{cm}^2\text{s})\), much less than that of copper \((5 \times 10^{-9}\text{m}^2\text{s})\) [22], the precipitated particles of Tₘₐₙ and AlCu₃Mn₂ phase are fine and dispersed. With Cu content in the alloy, the area of the Cu-rich regions is increased, and hence, the area of PZs is also increased. In the alloy with high Cu content, the degree of supersaturation becomes larger, and thus, the driving force for precipitation becomes stronger that leads to an increase in amount and a decrease in size of the precipitated particles. But in the core part of dendrites, the Cu solute is much poor, so no precipitation occurs, forming NPZs in the original core parts of dendrites.

**Figure 3:** Transmission electron micrographs of precipitated particles in A₁ (a), A₂ (e), and A₃ (i) alloys as-solutionized; representative EDS (b) and SAD (c) results of the Tₘₐₙ phase in the A1 alloy; EDS (d) and SAD (g) results of the AlCu₃Mn₂ phase and EDS (f) of the Tₘₐₙ phase in the A2 alloy, respectively, marked by Arrows 1 and 2 in (e); EDS results (h, j) of the particles AlCu₃Mn₂ and Tₘₐₙ in the A3 alloy, respectively, marked by Arrows 3 and 4 in (i).
Figure 4: Optical microstructures of three Al-Cu-Mn alloys as-cast: (a) $A_1$ alloy; (b) $A_2$ alloy; (c) $A_3$ alloy.

Figure 5: Continued.
As seen in Figure 1, a number of PFBs are formed, and it becomes more apparent with Cu content in the alloy. Figure 2 more clearly exhibits a PFB between PZs, marked by Arrow 3. In the PFB, there are some CuAl$_2$ particles which survived. Arrow 2 in Figure 2 presents the blank region in PZs. Figure 9 illustrates the EDS results of the regions denoted by Arrows 2 and 3. In the blank region in PZs, the content of Mn is still high to about 0.7 wt.%, and even $T_{\text{Mn}}$ and AlCu$_3$Mn$_2$ particles have

**Figure 5:** SEM images of Al-Cu-Mn alloys as-cast: (a) A$_1$ alloy; (b) A$_3$ alloy; (c, d) EDS results of the $T_{\text{Mn}}$ and $\theta$ phases in the A$_1$ alloy, respectively; (e, f) EDS results of the $T_{\text{Mn}}$ and $\theta$ phases in the A$_3$ alloy, respectively.

**Figure 6:** SEM image showing the light-grey and dark-grey areas of dendrites in the A$_2$ alloy as-cast (a) and the EDS results of the light-grey area (b) and dark-grey area (c).
Figure 7: EPMA microstructure in the A₂ alloy as-cast, showing the distribution of solute elements of Cu, Fe, and Mn (for interpretation of the references to color in this figure legend, the reader is referred to the web version of this article).

Figure 8: SEM images showing the variation of eutectic phases after solution treatment: (a) A₁ alloy; (b) A₂ alloy; (c) A₃ alloy.
precipitated around it, but in the PFB, no Mn trace is detected. Figure 7 illustrates that the regions where the CuAl2 phase exists alone are highly Mn-poor. With Cu content in the alloy, the amount of the eutectic CuAl2 phase formed in form of nonequilibrium is increased (as seen in Figures 4 and 5). During the course of solutionizing at 525°C for 6 hrs + 535°C for 6 hrs, first, this eutectic CuAl2 phase in form of nonequilibrium has to be redissolved into matrix, and then, the remaining CuAl2 phase becomes granulated. This evolution of redissolution and granulation of the eutectic CuAl2 phase leaves the band area where it is highly Mn-poor. It is due to Mn-poor that no precipitation of T\text{Mn} and AlCu3Mn2 particles takes place there, and thus, PFBs are formed among the PZs. In the alloy with higher Cu content, the amount of the eutectic CuAl2 phase that would be dissolved is increased, and thus, the formed PFBs become more apparent. It is concluded that it is the redissolution and granulation of the eutectic CuAl2 phase that lead to the formation of PFBs between PZs.

3.2. Contribution to Tension Properties. For the A2 alloy, the tension mechanical properties at different states are as follows: as-cast, as-solutionized, and T6 that are labeled in Table 2. In case of the tension test at room temperature (25°C), solutionizing treatment results in a remarkable increase in both the UTS (ultimate tension strength) and YS (yield strength) by 74.1% and 74.4%, respectively, without sacrificing the elongation, and further aging treatment also considerably increases the UTS (from 256 MPa to 369 MPa), YS (from 157 MPa to 248 MPa), and elongation (from 4.8% to 6.3%) again. But, under the tension test at 300°C, the case is different. Solutionizing treatment produces an appreciable increase in UTS and YS by 29.2% and 36.8%, respectively; however, further aging treatment does not lead to obvious improvement of strength both in YS and UTS. Engineering stress verses strain curves shown in Figure 10 vividly exhibit the contribution of microstructure evolution to mechanical properties. The curves of tension at room temperature in Figure 10(a) are common as expected. Both the microstructure evolutions occurring during the solutionizing and aging courses contribute to their respective strengthening effects. Curves in Figure 10(b) reveal the physic process in strengthening at high temperature. In case of tension at 300°C, for the sample as-cast, hardening occurs with further plastic deformation beyond yielding, but it merely sustains for only about 4% strain in total, and then a collapse failure arrives suddenly. For samples as-solutionized and T6 tempered, the curves of the two samples are almost the same at the first half stage, with almost equal values both in UTS and YS. Surely, the values are much higher than that as-cast. And the hardening behavior also occurs with further plastic deformation beyond yielding. However, after the hardening reaches a maximum at about 4 w.% strain in total, slightly softening occurs. It continues to a much larger strain of about 8% and 10% in total, respectively, until the final collapse failure. This softening is from dynamic recovery or recrystallization during plastic deformation with a slow strain rate (about 0.0033 s\(^{-1}\)) at high temperature, which needs more attention. Compared with the sample as-cast, what support does the samples get (as-solutionized and T6) to tolerate much larger plastic deformation? In the A2 alloy as-cast, the microstructure is constituted with the dendritic \(\alpha\)-Al phase and small amount of the eutectic \(T\text{Mn}\) and \(\theta\) phases. After solutionizing, a great number of fine \(T\text{Mn}\) and AlCu3Mn2 dispersoid particles are precipitated in the A2 alloy except for the redissolving of part of the eutectic \(\theta\) phase (Figures 1–3 and 8). Once these \(T\text{Mn}\) and AlCu3Mn2 particles are formed during the solutionizing course, they will almost not change whether in morphology or in size even when they undergo another thermal history with a temperature less than the solutionizing temperature. It is due to its complicated structure and much low diffusion coefficient of Mn in the Al solution [12, 18, 23]. In the sample after aging treatment, the needle-like \(\theta\)′ and slender rod-like \(\theta\) phases in the conventional Al-Cu alloy have not been observed. Conversely, a number of very fine particles (dozens of nanometers in size) are found to precipitate on the rod-like \(T\text{Mn}\) particles, as shown in

<table>
<thead>
<tr>
<th>State</th>
<th>UTS (MPa)</th>
<th>YS (MPa)</th>
<th>Elongation (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>As-cast</td>
<td>25°C: 256 ± 18</td>
<td>90 ± 4.0</td>
<td>5.0 ± 0.7</td>
</tr>
<tr>
<td></td>
<td>300°C: 456 ± 18</td>
<td>226 ± 5.8</td>
<td>5.0 ± 0.6</td>
</tr>
<tr>
<td>As-solutionized</td>
<td>25°C: 256 ± 18</td>
<td>157 ± 1.6</td>
<td>4.0 ± 0.7</td>
</tr>
<tr>
<td></td>
<td>300°C: 456 ± 18</td>
<td>226 ± 5.8</td>
<td>5.0 ± 0.6</td>
</tr>
<tr>
<td>T6</td>
<td>25°C: 256 ± 18</td>
<td>157 ± 1.6</td>
<td>4.0 ± 0.7</td>
</tr>
<tr>
<td></td>
<td>300°C: 456 ± 18</td>
<td>226 ± 5.8</td>
<td>5.0 ± 0.6</td>
</tr>
</tbody>
</table>

Figure 9: EDS results of the regions marked by Arrow 2 (a) and Arrow 3 (b) in Figure 2(a).
Figure 11(a). Its EDS results (Figure 11(b)) indicate that it contains a large amount of Cu and Al and a small amount of Mn and Fe. However, due to its Cu/Al atom ratio close to 1:2, these fine precipitates may be the metastable $\theta$ phase. The precipitation behavior of these fine particles during aging may be responsible for the considerable increment in room temperature strength of the T6 samples compared with the as-solutionized samples. These particles precipitated during the aging course can strengthen the matrix in case of tension at room temperature. But usually, these particles formed during aging possess simple structure, and the diffusion coefficient of the constituents is high, such as Cu, Si, and Mg in the Al solution, and thus, these particles will be coarsened quickly even redissolved when they undergo another thermal history [24–27]. So, the particles of the Al-Cu-Mn phases and metastable $\theta$ phase precipitated during the solutionizing and aging courses, respectively, remarkably increase the YS and UTS at room temperature (Figure 10(a) and Table 2). But the microstructure evolution occurring in the aging course does not lead to an improvement in YS and UTS at 300°C (Figure 10(b)). It is a great number of $T_{Mn}$ and AlCu$_3$Mn$_2$ particles precipitated during solutionizing that support the samples (as-solutionized and T6) to tolerate much larger plastic deformation. That is, the considerable increments in YS and UTS at 300°C of the $A_2$ alloy as-solutionized and T6 tempered are almost completely contributed to the dispersoid precipitation of a great number of $T_{Mn}$ and AlCu$_3$Mn$_2$ particles. Meanwhile, it also indicates that the precipitation behavior in the aging course has no or little contribution to high temperature strength. On other words, if a component is working at a low temperature or room temperature, aging treatment is necessary. And if it is at high temperature, such as at/above 300°C, solutionizing treatment is enough, and aging treatment is completely unnecessary.
The tension mechanical properties at 25°C and 300°C of the studied alloys after T6 treatment are listed in Table 3. In case of tension at 25°C, the YS is increased significantly from 66 MPa to 303 MPa with Cu content in the alloy changing from 2.0 wt.% to 7.5 wt.%. But for UTS, the A2 alloy has the maximum of 369 MPa, with about 45 MPa higher than that of the A3 alloy. The A3 alloy contains 7.5 wt.% Cu, with a higher Cu content than the A2 alloy (4.5 wt. % Cu). In as-solutionized microstructure of the A3 alloy, the amount of the survived eutectic \( \theta \) phase at grain boundaries is much higher than that of the A2 alloy (Figure 8). During tension at 25°C, because the amount of the \( T_{\text{Mn}} \) and \( \text{AlCu}_3\text{Mn}_2 \) particles in the A3 alloy precipitated during solutionizing is larger than that in the A2 alloy and its size is also finer, the YS of the A3 alloy is higher than that of the A2 alloy (Table 3), but more severe grain boundary brittleness in the A3 alloy that is induced by the survived \( \theta \) phase at grain boundaries makes the matrix lose the ability to tolerate large plastic deformation at room temperature; thus, only with a smaller strain, the sample is fractured. As a result, the elongation of the A3 alloy (about 1.1%) is much less than that of the A2 alloy (about 6.3%), and hence, the UTS is also decreased. In case of tension at 300°C, both the YS and UTS are increased with Cu content due to the increased precipitation hardening with the Cu content, consistent with the results in [28]. As concluded above, high temperature strength overridingly depends on the precipitation behavior during solutionizing. The number of particles precipitated during solutionizing is increased with Cu content in the alloy, and simultaneously, the particle size is decreased (Figure 3); the survived eutectic \( \theta \) and \( T_{\text{Mn}} \) particles at the grain boundaries can also enhance the ability of grain boundaries to oppose plastic deformation at high temperature, and thus, the high temperature strength is considerably increased with Cu content in the alloy.

### Data Availability

The data used to support the findings of this study are included within the article.

### Conflicts of Interest

The authors declare that they have no conflicts of interest.

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