Original Article

Thermal exposure of Al-Si-Cu-Mn-Fe alloys and its contribution to high temperature mechanical properties

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\section*{A B S T R A C T}

The effect of Cu and Ti contents and thermal exposure on the elevated temperature mechanical properties of T6-treated Al-Si-Cu-Mn-Fe alloys was investigated via optical microscopy (OM), scanning electron microscope (SEM), transmission electron microscope (TEM), and elevated temperature tensile tests. The mechanical properties at elevated temperature increased with the increase in Ti and Cu contents. These results can be attributed to the increased amounts of precipitated particles. Thermal exposure at 300 °C for 10 h had a deleterious effect on the elevated temperature tensile properties due to the coarsening of the strengthening precipitates. Further thermal exposure up to 100 h only resulted in a little reduction in the strength. After thermal exposure at elevated temperature of 300 °C, the 4Cu-0.5Ti demonstrated best mechanical properties and Quality Index (QI). The ultimate tensile strength (UTS) at elevated-temperature of 300 °C were 170 MPa, 92 MPa, and 88 MPa for 0.5 h, 10 h, and 100 h after thermal exposure, respectively, which are superior to those for the commercial aluminum alloys.

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\section*{1. Introduction}

In order to promote the development of low-carbon economy for energy-saving and emission-reduction, the demands for high compression and engine efficiency in automotive engines are increasing. Therefore, it is necessary to improve the heat resistant properties of aluminum alloys for the applications in diesel oil automotive engine components for military and marine industries. The high temperature thermal exposure experiment of aluminum alloys is commonly used in the automotive engines to represent the real service conditions. The elevated temperature thermal exposure may result in some microstructure changes, such as formation of novel phases, coarsening of precipitates and intermetallics, development of precipitate free zones, and variation of dislocation and microcracks [1,2]. Hence, the microstructural changes after thermal exposure would affect the tensile properties.

Al-Si cast alloys are widely employed as key engine components in the automotive industry owing to their superior casting properties, low density, good corrosion resistance, and high wear and heat resistance [3]. The Al-Si-Mg cast alloys,
such as A356 and A357 alloys, have been found to have some limitations when used as engine components below the temperature of 200 °C [4–7]. Ceschin studied the effect of thermal exposure on mechanical properties of the A356 cast aluminum alloy. The UTS and YS of A356-T6 alloy decreased 32 % and 40 %, respectively, after thermal exposure at 210 °C for 41 h. The decrease in strength can be attributed to the rapid coarsening of the precipitates [4]. The Al-Si-Cu-Mg alloys can be used as heat resistant components at elevated temperature of 250 °C over long periods of time because of the formation of Al2Cu and W (AlxMgsCuxSi) phases [8–11]. Ceschin also explored the effect of thermal exposure on the mechanical properties of C355 alloy (Al-Si-Cu-Mg) after thermal exposure at 210 °C for 41 h and showed that the tensile properties of C355 alloy were better than the A356 alloy [9].

Adding the transition element, such as Sc, Zr, Ni, and V, to Al-Si-Mg and Al-Si-Cu-Mg alloys can improve the thermal exposure mechanical properties [5–8]. Ding concluded that the improvement of the tensile properties and the thermal stability of the A356 alloys under thermal exposure can be attributed to the nano Al2Sc phase, which allows these alloys to be used as engine components up to 200 °C [6]. Tzeng studied the effect of Zr on the thermal exposure mechanical properties of A357 at 250 °C and found that Zr addition clearly improved the thermal exposure mechanical properties owing to the formation of Al2Zr precipitates during the thermal exposure process [7]. Abdelaziz explored the thermal exposure of Al-Si-Cu-Mg alloys with the addition of Ni, Zr, and Mn and concluded that the addition of Ni or Mn to the Zr-containing Al-Si-Cu-Mg cast alloys has a little effect on the mechanical properties of the alloys after thermal exposure [8]. For the Al-Si-Cu-Ni alloys, it was reported that the increasing volume fraction of Al2CuNi phases was responsible for the improvement of the thermal exposure mechanical properties at 350 °C for 200 h [12]. For the Al-Si-Cu-Mg-Ni alloys, the thermal exposure at 350 °C and 420 °C deteriorated the mechanical properties at both room and elevated temperatures. However, the exposure mechanical properties at 420 °C were better than that at 350 °C, which can be attributed to the formation of nano sized Al2CuMn1S3 phase [13].

In recent years, Mn and Fe elements have been added to the heat resistant Al-Si alloys to improve the elevated temperature mechanical properties. This phenomenon can be attributed to the formation of heat resistant iron-rich and Mn-rich intermetallics [14–17]. Wang et al. found that the formation of Chinese script iron-rich intermetallics was beneficial for enhancing the mechanical properties of Al-Si-Cu-Mn-Fe alloys [14]. Liu et al. studied the mechanical properties of Al-13 % Si piston alloys with Mn addition at elevated temperatures and discovered that both the yield strength (YS) and creep resistance at 300 °C were improved, which can be attributed to the heat resistant α-(MnFe) iron-rich intermetallics [15]. Liao et al. developed an Al-Si-Cu-Mn heat-resistant piston alloy and found that the tensile strength of the alloy at elevated temperature was improved due to the formation of Chinese script Mn-rich intermetallics [16]. Suo et al. concluded that there was no change in the morphology and size of the Mn-rich phase after the thermal treatment at 525 °C for 20h, indicating that Mn-rich phase had excellent thermo-stability [17]. Therefore, the addition of alloying elements such as Fe and Mn is an efficient method for developing recycled aluminum alloys used in the industries that demand heat-resistant components.

However, all these works have focused on the instantaneous tensile properties of Al-Si-Cu-Mn-Fe alloys at elevated temperatures [14–17]. The elevated temperature thermal exposure experiment of Al-Si-Cu-Mn-Fe alloys has seldom been reported. Besides, we also examined the effect of Ti content on the microstructure and mechanical properties of Al-Si alloys in the as-cast conditions. These results clearly indicate that the TiAlSi phase formation and grain refinement improved the alloy mechanical properties [18]. Additionally, the TiAlSi phase is considered to be beneficial for the improvement of mechanical properties at elevated temperatures [19]. In the present work, the effect of Cu and Ti contents and thermal exposure on the elevated temperature mechanical properties of heat-treated Al-Si-Cu-Mn-Fe alloys was investigated. This study would be beneficial to describe the behavior of the Al-Si-Cu-Mn-Fe alloys during service in real applications, such as engine components in the military and marine industries.

2. Materials and methods

2.1. Alloy melting

Commercially pure Al (99.5 %) and master alloys (such as Al-20 % Si, Al-50 % Cu, Al-10 % Mg, Al-10 % Mn, Al-5 % Fe, Al-20 % Sr, and Al-5 % Ti, all compositions quoted in this work are in weight percent unless indicated otherwise) were employed to prepare the experimental alloys with different Cu and Ti amounts. The alloy compositions were determined using optical emission spectrometry (OES), and the results are shown in Table 1. The melt temperature was maintained at 730 °C for 30 min. Approximately, 10 kg of melt was degassed using argon to reduce the hydrogen content. The melt was then poured into a cylindrical mold of size 80 mm in height and 50 mm in diameter. The pouring temperature and die temperature were set at 730 °C and 250 °C, respectively.

2.2. T6 heat treatment and tensile test

All samples for tensile test were cut into the dimension of Φ 10 mm × 80 mm by line-cutting machine from the same radius of the castings. The T6 heat treatment was used in this study in order to stabilize the microstructure [14]. The samples were solution treated at 505 °C for 8 h, and then, quenched in warm water at 100 °C. The samples were then aged at 160 °C for 12 h. The heat-treated test bars were then thermally exposed at 300 °C for different times (0.5h, 10h, and 100h). Tensile properties at elevated temperatures were obtained according to our previous work [20]. Tensile tests were carried out using an MTS CMT5105 standard testing machine at 300 °C with a holding time of 30 min in a constant-temperature box. The heating rate was 15 °C/min, while the extension rate was 2.0 mm/min. And the gauge length of the tensile test bars was 25 mm. The reported values were the average of at least three samples.
Table 1 – Chemical composition of the alloys.

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Si</th>
<th>Cu</th>
<th>Mn</th>
<th>Fe</th>
<th>Ti</th>
<th>Mg</th>
<th>Sr</th>
<th>Al</th>
</tr>
</thead>
<tbody>
<tr>
<td>6Cu-0Ti</td>
<td>6.72</td>
<td>5.96</td>
<td>0.61</td>
<td>0.63</td>
<td>0.028</td>
<td>0.13</td>
<td>0.03</td>
<td>Balance</td>
</tr>
<tr>
<td>6Cu-0.2Ti</td>
<td>6.91</td>
<td>5.88</td>
<td>0.61</td>
<td>0.60</td>
<td>0.16</td>
<td>0.14</td>
<td>0.03</td>
<td>Balance</td>
</tr>
<tr>
<td>6Cu-0.5Ti</td>
<td>6.76</td>
<td>5.91</td>
<td>0.60</td>
<td>0.62</td>
<td>0.45</td>
<td>0.12</td>
<td>0.03</td>
<td>Balance</td>
</tr>
<tr>
<td>4Cu-0Ti</td>
<td>6.82</td>
<td>3.86</td>
<td>0.58</td>
<td>0.61</td>
<td>0.027</td>
<td>0.14</td>
<td>0.03</td>
<td>Balance</td>
</tr>
<tr>
<td>4Cu-0.2Ti</td>
<td>6.72</td>
<td>3.90</td>
<td>0.39</td>
<td>0.59</td>
<td>0.22</td>
<td>0.12</td>
<td>0.03</td>
<td>Balance</td>
</tr>
<tr>
<td>4Cu-0.5Ti</td>
<td>6.93</td>
<td>3.85</td>
<td>0.62</td>
<td>0.63</td>
<td>0.48</td>
<td>0.15</td>
<td>0.03</td>
<td>Balance</td>
</tr>
</tbody>
</table>

where $d$ is the material constant (150 for the Al-7Si-Mg alloys in the room temperature [5]). Till date, there is no available published data on $d$ at elevated temperatures. As a result, the $d$ value was taken as 230, which was calculated by using MATLAB software based on the tensile results from Fig. 1. As shown in Fig. 1d, the 4.0Cu-0.5Ti alloy had the best QI value. The QI values were 311, 300, and 303 MPa for 0h, 10h, and 100h after thermal exposure at 300 °C. The best QI value for 4.0Cu-0.5Ti alloy can be attributed to the higher UTS and elongation.

2.3. Microstructure observation

The samples for metallographic observation were cut from the gauge length part of the selected tensile specimens and were etched in Keller solution for 10 s. The morphology, energy dispersive spectrum (EDS) of the secondary intermetallics phases, and surface fractures were analyzed under SEM (Nova Nano SEM 430). The precipitates in the $\alpha$(Al) matrix were analyzed using a TEM (JEOL JEM-3010) at 200 kV. The volume fraction of the dispersoids $f$ was calculated using the following equation [21]:

$$f = A_h - \frac{K_D}{K_D + t}(1 - A_{DFZ})$$

(1)

where $D$ is the average equivalent diameter of dispersoids in the TEM images, $A_h$ is the area fraction of dispersoids in the TEM images, $A_{DFZ}$ is the volume fraction of dispersoids free zone (DFZ) measured in the optical images, $t$ is the TEM foil thickness, and $K$ is the average shape factor of the dispersoids.

3. Results and discussion

3.1. Mechanical properties of Al-Si-Cu-Mn-Fe alloys

Fig. 1 shows the elevated temperature mechanical properties of Al-Si-Cu-Mn-Fe alloy after thermal exposure at 300 °C. The thermal exposure time clearly deteriorated the elevated temperature mechanical properties. The thermal exposure mechanical properties at 300 °C for 0.5 h still had high strength values. Further prolonged exposure at 300 °C for 10 h had a deleterious effect on the elevated temperature tensile properties. Further thermal exposure up to 100 h only resulted in a little reduction in the strength. For instance, for the elevated-temperature ultimate tensile strength of 6.0Cu-0.5Ti alloys, the values were 182, 99, and 90 MPa for 0.5 h, 10 h, and 100 h after thermal exposure at 300 °C. The elevated temperature mechanical properties increased with the increase in Ti and Cu contents after thermal exposure. However, the increase in elevated temperature mechanical properties with Ti and Cu addition decreased with the rise in thermal exposure duration.

To further evaluate the effect of Cu and Ti contents on the elevated temperature mechanical properties of Al-Si-Cu-Mn-Fe alloys after thermal exposure, the QI was calculated in this work (see Fig. 1d). The QI is a comprehensive tensile properties parameter which is related to the combination of UTS and elongation (EL) [22]. The QI of the Al-7Si-Mg alloys can be calculated by Eq. (2):

$$QI = UTS + d \log EL$$

(2)

Fig. 2 shows the heat resistant intermetallics in the Al-Si-Cu-Mn-Fe alloys with different Cu and Ti contents. The SEM-EDS results of the secondary intermetallics are presented in Table 3. As per the EDS results of Table 3, the secondary intermetallics were white block $\alpha_2$Cu, Chinese script iron-rich intermetallics $\alpha$-Fe, and the needle-like intermetallics $\alpha_3$Cu$_2$Fe and TiAlSi, which are similar to the previous reports [19,23–26]. As shown in Fig. 2a, primarily white block $\alpha_2$Cu and dark gray Chinese script $\alpha$-Fe were present in the 4.0Cu-0.5Ti alloy. In the 4.0Cu-0.2Ti alloy, the second intermetallics were primarily $\alpha_2$Cu and $\alpha$-Fe, while the needle-like phases were occasionally detected in the sample Fig. 2b. Similarly, a few needle-like secondary intermetallics were also observed in the 4.0Cu-0.5Ti alloy, apart from the $\alpha_2$Cu and $\alpha$-Fe (Fig. 2c). Compared with 4.0 % Cu alloys, a large amount of $\alpha_2$Cu was observed in the 6.0 % Cu alloys because of the supersaturation of Cu solid solubility (Fig. 2d–f). With the increase in Ti content, the microstructures and morphologies of the two alloys with different Cu contents were almost the same. The needle-like intermetallics clearly increased, but $\alpha_2$Cu and Chinese script $\alpha$-Fe decreased significantly with the Ti addition.

Fig. 3 shows the microstructures of the Al-Si-Cu-Mn-Fe alloys after exposure at 300 °C for 0.5 h, wherein, the addition of Cu and Ti clearly increased the amount of $\theta$ ($\alpha_2$Cu). The size of $\theta$ ($\alpha_2$Cu) in the alloys with different Cu and Ti contents were quantitatively analyzed by using Image-Pro Plus software. The size of $\theta$ phases was 190 nm in 4Cu-0Ti alloy, which then decreased to 110 nm in the 6.0Cu-0Ti alloy, finally decreasing to 90 nm in the 6.0Cu-0.5Ti alloy. The micrometer-
sized precipitation particles were the $T\,(Al_{20}Cu_3Mn_3)$ phase in the Al-Cu-Mn and Al-Si-Cu-Mn alloys [16,27]. While, the effect of Cu and Ti contents on the amount of $T\,(Al_{20}Cu_3Mn_3)$ had a little change. The amount of $\eta$ phases in the alloys increased with the increase in Ti content, which can mainly be attributed to the grain refinement [28]. The grain size can affect the amount and size of precipitates in the Al-Cu alloys [29,30], which is because the refined grains lead to a shorter diffusion distance of Cu atom during solution treatment [31].

Fig. 4 shows the microstructures of the Al-Si-Cu-Mn-Fe alloys after exposure at 300 °C for 100 h. The nanoscale blocky phase were $\alpha$-Fe(Al$_{15}$(FeMn)$_3$Si$_2$) phase, which have been proved to exist in Al-Cu-Mn and Al-Si-Cu-Mn alloys [15,16,27]. The size and amounts of $T\,(Al_{20}Cu_3Mn_3)$ and

Table 2 – Mechanical properties of the Al-Si-Cu-Mn-Fe alloys and previously reported alloys at the elevated temperature thermal exposure [2].

<table>
<thead>
<tr>
<th>Alloys (Condition)*</th>
<th>Temperature (°C)</th>
<th>UTS (MPa) after various thermal exposure</th>
<th>Source</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>0.5 h</td>
<td>10 h</td>
</tr>
<tr>
<td>4Cu-0Ti(T6)</td>
<td>300</td>
<td>132</td>
<td>77</td>
</tr>
<tr>
<td>4Cu-0.5Ti(T6)</td>
<td>300</td>
<td>170</td>
<td>92</td>
</tr>
<tr>
<td>6Cu-0Ti(T6)</td>
<td>300</td>
<td>145</td>
<td>82</td>
</tr>
<tr>
<td>6Cu-0.5Ti(T6)</td>
<td>300</td>
<td>182</td>
<td>99</td>
</tr>
<tr>
<td>2024(T6)</td>
<td>315</td>
<td>110</td>
<td>85</td>
</tr>
<tr>
<td>6061(T6)</td>
<td>315</td>
<td>85</td>
<td>62</td>
</tr>
<tr>
<td>7075(T6)</td>
<td>315</td>
<td>70</td>
<td>62</td>
</tr>
<tr>
<td>356.0(T6)</td>
<td>315</td>
<td>52</td>
<td>45</td>
</tr>
<tr>
<td>319.0(T6)</td>
<td>315</td>
<td>90</td>
<td>75</td>
</tr>
</tbody>
</table>

Table 3 – Average composition of the secondary intermetallics, at %.

<table>
<thead>
<tr>
<th>Source</th>
<th>Phase</th>
<th>Al</th>
<th>Cu</th>
<th>Mn</th>
<th>Fe</th>
<th>Si</th>
<th>Ti</th>
</tr>
</thead>
<tbody>
<tr>
<td>6Cu-0.5Ti</td>
<td>$Al_2Cu$</td>
<td>69.60</td>
<td>29.20</td>
<td>–</td>
<td>–</td>
<td>1.2</td>
<td>–</td>
</tr>
<tr>
<td>6Cu-0.5Ti</td>
<td>$\alpha$-Fe</td>
<td>70.85</td>
<td>1.81</td>
<td>7.00</td>
<td>3.34</td>
<td>2.84</td>
<td>7.79</td>
</tr>
<tr>
<td>6Cu-0.5Ti</td>
<td>$Al_2Cu_2Fe$</td>
<td>73.65</td>
<td>12.38</td>
<td>3.34</td>
<td>2.84</td>
<td>7.79</td>
<td>–</td>
</tr>
<tr>
<td>6Cu-0.5Ti</td>
<td>TiAlSi</td>
<td>66.04</td>
<td>6.66</td>
<td>–</td>
<td>–</td>
<td>–</td>
<td>–</td>
</tr>
</tbody>
</table>

Fig. 1 – Elevated temperature mechanical properties of Al-Si-Cu-Mn-Fe alloy after thermal exposure: (a) UTS; (b) YS; (c) EL; (d) QI.
\[ \alpha-\text{Fe} \ (\text{Al}_{15}(\text{FeMn})_3\text{Si}) \] had little change. The size and volume percent of dispersoids in the alloys with different Cu and Ti contents were quantitatively analyzed (Fig. 5). The T (\text{Al}_{20}\text{Cu}_2\text{Mn}) and \( \alpha \)-Fe (\text{Al}_{15}(\text{FeMn})_3\text{Si}) sizes in the 4Cu-0Ti alloy were estimated to be 348 nm and 114 nm, respectively; in the 6Cu-0Ti alloy, the same were 302 nm and 98 nm, respectively; whereas in the 6Cu-0.5Ti alloy, the same were 276 nm and 75 nm, respectively.

Fig. 6 shows the fractured surfaces of Al-Si-Cu-Mn-Fe alloys with different Cu and Ti contents after exposure at 300 °C for 0.5 h. As shown in Fig. 6a, several dimples were observed in the fractured surfaces of the 4Cu-0Ti alloy, indicating that the alloys displayed the characteristics of a ductile fracture. In 6Cu-0Ti alloy, mainly dimples and cleavage step were found in the fractured surfaces, suggesting that the alloy demonstrated ductile and brittle hybrid failure characteristics. These results signify that the Cu addition in Al-Si-Cu-Mn-Fe alloy deteriorated the ductile properties, which can be attributed to the formation of excessive \( \beta \)-Cu phases (Fig. 6b). For the alloys with Ti addition (Fig. 6c and d), there were some needle-like intermetallics and cracks in the fractured surfaces except for some dimples and cleavage step, which indicate that these alloys presented worse brittle failure characteristics as compared to the free Ti addition alloys. The brittle failure characteristics of the 4Cu-0.5Ti and 6Cu-0.5Ti alloys can be attributed to the formation of needle-like \( \beta \)-Fe (\text{Al}_{15}\text{Cu}_2\text{Fe}) and TiAlSi intermetallics. The needle-like intermetallics are often present as weak locations in the matrix and facilitate the formation of cracks during the tensile process, which can be clearly observed in the fracture surfaces. Therefore,
both Cu and Ti addition in the Al-Si-Cu-Mn-Fe alloy would decrease the ductility to a certain extent.

Fig. 7 illustrates the longitudinal microstructure beneath the fractured surfaces of Al-Si-Cu-Mn-Fe alloys with different Cu and Ti contents after exposure at 300 °C for 0.5 h. A large amount of Si particles and α-Fe were present at the fractured surface in the 0 % Ti alloys with different Cu contents, indicating that the fracture happened due to the presence of eutectic Si particles and α-Fe (Fig. 7a and b). A large amount of eutectic Si particles, needle-like phase precipitated at or beneath the fractured surface of the 0.5 % Ti alloys with different Cu contents; thus, resulting in crack propagation and fracture (Fig. 7c and 7d). These results further prove that the Ti addition affects the fracture type of the Al-Si-Cu-Mn-Fe alloys due to the formation of needle-like intermetallics. While, the Cu addition has a little influence on the fracture type during elevated temperature tensile loading may because of the softening of Al2Cu at high elevated temperatures [14].

3.3. Microstructures of Al-Si-Cu-Mn-Fe alloys after elevated temperature thermal exposure for various times

Fig. 8 shows the dispersoid morphology in the α-Al matrix of 4Cu-0.2Ti alloys after elevated temperature thermal exposure for various times. The precipitated particles can be clearly observed in the OM images which were etched by Keller reagent (Fig. 8a–c). The precipitated particles clearly decreased with the increase in the thermal exposure duration. However, upon further increase in the exposure duration from 10 h to 100 h, the precipitated particles almost had no change. As shown in Fig. 8d–f, in order to reveal the change in precipitated particles after various thermal exposure times, TEM analysis was used. The precipitated particles coarsened with the increase in thermal exposure times at elevated temperature. As the thermal exposure time increased above 10 h, the amount of fine needle-like α phases clearly decreased. Upon further increase in thermal exposure up to 100 h, the amount and size of T (Al20Cu3Mn3) and α-Fe (Al15(FeMn)3Si2) in alloys had a little change. The T (Al20Cu3Mn3) and α-Fe are the most common heat-resistant phases in aluminum alloys, and the heat resistance temperature can be as high as 350–400 °C [16,27]. Therefore, the Al-Si-Cu-Mn-Fe alloys have excellent mechanical properties at elevated temperatures after thermal exposure.

Fig. 9 illustrates the fracture surfaces and microstructure beneath the fracture surfaces of 4Cu-0.2Ti alloys after elevated temperature thermal exposure for various times. There were several dimples in the fractured surfaces, which indicate that these alloys had the characteristics of the ductile fracture. As the thermal exposure time increased up to 10 h, the
Fig. 6 – Fracture surfaces of the Al-Si-Cu-Mn-Fe alloys after exposure at 300 °C for 0.5 h: (a) 4Cu-0Ti; (b) 6Cu-0Ti; (c) 4Cu-0.5Ti; (d) 6Cu-0.5Ti.

Fig. 7 – Longitudinal microstructure beneath the fracture surfaces of Al-Si-Cu-Mn-Fe alloys after exposure at 300 °C for 0.5 h: (a) 4Cu-0Ti; (b) 6Cu-0Ti; (c) 4Cu-0.5Ti; (d) 6Cu-0.5Ti.
4. Conclusions

The effect of Cu and Ti contents and the thermal exposure at 300 °C on the elevated temperature mechanical properties of heat-treated Al-Si-Cu-Mn-Fe alloys was investigated. The conclusions are as follows:

1. The elevated temperature mechanical properties increased with the increase in Ti and Cu contents after thermal exposure. These results were attributed to the increase in amounts of the precipitated particles, including $\theta$ (Al$_2$Cu), T (Al$_{20}$Cu$_2$Mn$_3$), and the $\alpha$-Fe(Al$_{13}$FeMn)$_3$Si$_2$. However, the enhancement in elevated temperature mechanical properties with Ti and Cu addition decreased with the increase in the thermal exposure time, which can be attributed to the softening of $\theta$ (Al$_2$Cu) and grain refinement.

2. The thermal exposure mechanical properties at 300 °C for 0.5 h still had high strength value, which can be attributed to the amounts of nano sized $\theta$ (Al$_2$Cu). The coarsening of the strengthening precipitates following a prolonged exposure at 300 °C for 10 h had a deleterious effect on the elevated temperature tensile properties. Further thermal exposure up to 100 h resulted in little reduction in
the strength, which can be attributed to the high heat resistance of $T$ (Al$_2$Cu$_2$Mn$_2$) and the α-Fe (Al$_{15}$(FeMn)$_3$Si$_2$) precipitated particles.

(3) After thermal exposure at 300 °C, the heat-treated Al-Si-Cu-Mn-Fe alloy with 4.0 % Cu and 0.5 % Ti demonstrated the best comprehensive mechanical properties and the quality indices. The values of elevated-temperature ultimate tensile strength were 170 MPa, 92 MPa, and 88 MPa for 0 h, 10 h, and 100 h of thermal exposure, respectively, which are superior to those of the commercial aluminum alloys. The best quality indices can be attributed to the high strength and ductility.

Conflict of interest

The authors declare no conflicts of interest.

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