Investigation of the Effect of Heat Treatment on the Microstructures and Mechanical Properties of Al-13Si-5Cu-2Ni Alloy

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Abstract: An experimental investigation was carried out to study the effects of solid solution treatment and aging treatment on the microstructures and mechanical properties of an Al-13Si-5Cu-2Ni alloy. The results show that the size of eutectic silicon decreased with solid solution treatment temperature increasing until 510 °C. Subsequently, the eutectic silicon size continued to increase as the temperature increased to 520 °C. Initially, the acicular eutectic silicon of the as-cast alloy was 10.1 μm in size. After the solid solution treatment at 510 °C, the eutectic silicon size was reduced to 6.5 μm. The θ′ phase is the main strengthening phase in the alloy, therefore, the effect of aging treatment on θ′ phases was explored. As the aging time increased, the diameter, length, and fraction volume of the θ′ phases were found to increase. The main reason for the improved performance of this alloy following heat treatment is the passivation spheroidization of the silicon phase and Orowan strengthening due to the θ′ phases. The optimal tensile strength of an Al-13Si-5Cu-2Ni alloy was obtained after solid solution treatment at 510 °C for 8 h followed by an aging treatment at 165 °C for 8 h. Therefore, this work has great significance for promoting the application of Al alloys at high temperatures.

Keywords: Al-Si alloys; microstructure; solution treatment; ageing treatment

1. Introduction

With the increasing demand for light alloys, the use of Al-Si alloys is also increasing [1–3]. However, traditional binary Al-Si alloys are limited in application because of their lower strength and plasticity. The alloying of Al-Si alloys has also become an important research direction [4–6]. Researchers have developed various Al-Si alloys [7], such as Al-Si-Cu [8], Al-Si-Mg [9–13], Al-Si-Ni [14], Al-Si-Cu-Mg [15], and Al-Si-Cu-Ni [16] alloys. With the addition of multiple elements, the majority of intermetallic phases, such as Al2Cu, Mg2Si, Al3Ni, Al3CuNi, and Al7Cu4Ni, are formed [17]. The mechanical properties are improved with the formation of these intermetallic phases [18].

Adding Ni to Al-Si-Cu alloys will form various Ni phases, such as the Al3Ni phase, and these Ni phases will improve the elevated-temperature performance of such alloys; therefore, Al-Si-Cu-Ni alloys are usually applied to produce lightweight and high-performance pistons [19–21]. Researchers have performed many investigations for the heat-resistant Ni phase. Farkoosh [22] studied the effect of Ni additions on Al-7Si-0.5Cu-0.3Mg alloys. He found that the ratio of N to Cu can determine which precipitate is formed within the alloys. The γ-Al2Cu4Ni phase forms at Ni:Cu < 0.25; when Ni:Cu > 0.25, the δ-Al3CuNi phase forms, and, at Ni:Cu > 1.5, the formation of ε-Al3Ni phase occurs most likely. The γ-Al2Cu4Ni and δ-Al3CuNi phases can be dissolved into the matrix by solid solution
treatment, and these phases can be formed during aging treatment at temperatures above 300 °C. The heat treatment temperature and time can affect the amount and sizes of the precipitated phases [23].

Heat treatment can affect many aspects of metal materials, such as microstructure, the distribution of alloy elements, and mechanical properties [3,24–26]. Thus, the required microstructure and properties can be obtained through heat treatment processes [27,28]. Compared with other processes that are used to improve material properties such as alloying and modification, heat treatment has the advantages of low cost, simple operation, and significant effects. The alloy elements can be dissolved in the matrix by solid solution treatment. Strengthening phases can precipitate during the aging treatment, which significantly improves the strength of alloys [29]. In addition, solid solution treatment can fragment and spheroidize flake-like eutectic silicon. Therefore, solid solution treatment effectively reduces the stress concentration and wake crack propagation for Al-Si alloys. For Al-Cu alloys, a partial second phase (such as the Al$_2$Cu phase) can dissolve into α-Al grains during solid solution treatment, which leads to solid solution strengthening; fine and distributed uniform phases are present during the aging treatment, which results in aging strengthening [30,31]. Béroual et al. [32,33] found that the Cu element can be present in different shapes, such as compact blocky θ-Al$_2$Cu, in a eutectic with α-Al + θ-Al$_2$Cu phases or as a mixture of both types. Solid solution heat treatment at 525 °C maximizes the dissolution of θ-Al$_2$Cu. During the solid solution treatment, eutectic silicon fragments and spheroidizes first, following which the eutectic silicon coarsens with the increasing solid solution treatment time. The smaller size of eutectic Si reduces the damage of the Si phase dissevering the metal matrix. This improves the tensile strength and fatigue performance of the materials. For example, when the size of the eutectic Si particles is the smallest, the ultimate tensile strength will reach the highest point. With the increase in solid solution time, the coarsening of the eutectic Si phase and the dissolution of Al$_2$Cu in the matrix leads to a decrease in the ultimate tensile strength [34].

Temperature and time are two important parameters of heat treatment. It is necessary to choose an appropriate solid solution temperature. As the solid solution treatment temperature increases, the rate at which the Al$_2$Cu and Al-Ni phases dissolved in the matrix increases, which can result in alloy burnout and coarsening. Wang et al. [35] found that the melting point of multiple eutectic phases of AlSi$_{11}$Cu$_4$Mg$_{0.3}$ is 507 °C. Narayanan et al. [36] found that the Al$_2$Cu phase at the grain boundary of AlSi$_{6.0}$Cu$_{3.5}$Fe$_{1.0}$ alloys begins to melt at the range of 500–520 °C. As a consequence, combined with a solid solution treatment time of 8 h, an aging treatment temperature of 165 °C and an aging treatment time of 10 h, a temperature range of 480–520 °C was selected for solid solution treatment. Heat treatment is a simple and effective method to improve the mechanical properties of alloys. Therefore, it is important to choose an appropriate heat treatment process for Al-Si-Cu alloys. In short, heat treatment can influence the size of eutectic Si, the amount of Al$_2$Cu and Ni-Al intermetallic phases, and the mechanical properties of the alloys [37]. For heat treatable aluminum alloys, the appropriate heat treatment process has a great impact on their comprehensive mechanical properties [38].

At present, many studies have been performed on the heat treatment process of Al-Si-based alloys. Al-13Si-5Cu-2Ni alloy are high copper and high nickel alloys that are applied extensively for production and have high strength at elevated temperatures. It is meaningful to explore the influence of the T6 heat treatment process on the microstructure and mechanical properties of Al-13Si-5Cu-2Ni alloy. This work also provides a theoretical basis for the heat treatment process of Al-Si-Cu alloys. In the present work, the heat treatment process for Al-13Si-5Cu-2Ni alloy is studied to optimize the heat treatment process, which can provide theoretical and data support for the application of these alloys.
2. Experimental Procedure

2.1. Fabrication and Heat Treatment of Al-13Si-5Cu-2Ni Alloy

The materials used in this research were commercial pure Al ingots, Al-17Si alloys, copper scraps, and nickel, and the purity of all materials used was 99.99%. Al-13Si-5Cu-2Ni alloy was fabricated by steel mold casting. The chemical composition of the Al-Si-Cu alloys (wt.%) is shown in Table 1.

Table 1. Chemical composition of the Al-Si-Cu alloys (wt.%).

| Composition | Si   | Cu   | Ni   | Fe  | Al   |
|-------------|------|------|------|-----|------|
| Content (wt.%) | 12.87 | 5.03 | 1.94 | 0.16 | Balance |

Castings were subjected to T6 heat treatment under vacuum condition. A temperature range of 480–520 °C was selected for the solid solution treatment. Samples of the Al-13Si-5Cu-2Ni alloy were solution treated in a furnace at five different temperatures: 480, 490, 500, 510, and 520 °C for 8 h and then aging treated at 165 °C for 10 h. The microstructure and mechanical properties of these samples were tested, and the optimal temperature for solid solution treatment was chosen. The aging time was optimized based on the best solution treatment, and the aging treatment time ranged from 2 to 14 h.

2.2. Characterization

The phase compositions of the as-cast Al-13Si-5Cu-2Ni alloy were identified using X-ray diffraction (XRD). XRD analyses were performed with a D/Max 2500PC Rigaku diffractometer using Cu Kα radiation (400 kV/250 mA) at a scan rate of 4°/min in the range from 20° to 80°. As-cast and heat-treated samples were inspected using a scanning electron microscope (SEM, Evol8, Oberkochen, Germany) equipped with an energy-dispersive X-ray spectroscopy (EDS) attachment.

2.3. Performance Testing

The standard tensile test was conducted using an MTS-810 materials testing system at a constant strain rate of $1 \times 10^{-4} \text{s}^{-1}$. At least three samples were tested for every treatment condition. The microhardness of each sample was measured using a Vickers hardness tester, with an applied load of 50 N and a duration of 10 s. Additionally, eight samples (aging treated from 0 to 14 h at 165 °C) were tested for microhardness. Mean values for the Vickers hardness were obtained by testing 10 points on a single sample and removing the maximum and minimum values. Figure 1 shows dimensions of tensile sample.

Figure 1. Dimensions of tensile sample (unit:mm).

3. Results

3.1. Microstructure of As-Cast Al-13Si-5Cu-2Ni Alloy

The microstructure of as-cast Al-13Si-5Cu-2Ni alloy and compositional analysis are shown in Figure 2. As shown, five phases can be observed in Figure 2a,b. Five different phases are marked as A, B, C, D, and E in Figure 2b, respectively. From Figure 2c, it can be observed that the microstructure of as-cast Al-13Si-5Cu-2Ni alloy consists of α-Al phase, Si phase, Al$_2$Cu phase, and Al$_7$Cu$_4$Ni phase. It is clearly observed that the coarse block-like phases with sharp edges and corners is the primary silicon phase, marked A, and the gray stripe phase is the eutectic silicon phase, marked B. C and D phases were analyzed by an energy-dispersive spectrometer (EDS). The atomic percentage of the C and D phases is
given in Table 2. The Al:Cu ratio of the C phase is 2.6. The elemental content of Al is higher than the actual value due to the impact of the α-Al matrix. The C phase can be identified as Al$_2$Cu phases. The main elements of the D phase are Al, Cu, and Ni. The white stripe phase can be determined from the literature [16] as the Al$_7$Cu$_4$Ni phase, which is marked D. The matrix is the α-Al phase, marked E.

![Figure 2](image-url)

**Figure 2.** Microstructure and phase analysis of as cast Al-13Si-5Cu-2Ni alloy: (a) low magnification microstructure, (b) local magnification in figure (a), (c) XRD pattern, (d) C compositional analysis by EDS, and (e) D compositional analysis by EDS.

| Element | Al  | Si  | Cu  | Ni  | Phase          |
|---------|-----|-----|-----|-----|----------------|
| C       | 70.25 | 2.7 | 26.67 | 0.39 | Al$_2$Cu       |
| D       | 76.27 | 1.03 | 13.51 | 9.19 | Al-Cu-Ni       |

3.2. Effect of Solid Solution Treatment on Microstructure and Tensile Properties of Al-13Si-5Cu-2Ni Alloy

The microstructures of the as-cast and solid treated Al-13Si-5Cu-2Ni alloy at different temperatures for 8 h are shown in Figure 3. Figure 3a shows the typical cast microstructure of Al-13Si-5Cu-2Ni alloy. Block primary silicon with sharp edges and strip eutectic silicon is observed. The gray phases are the Al$_2$Cu and Al$_7$Cu$_4$Ni phases located at the grain boundaries, and the distribution of the second phases is reticular. For the solid solution treated at 480 °C (Figure 3b), the change in block-like primary silicon is not obvious, and the strip eutectic silicon is partially fragmented. The Al$_2$Cu and Al$_7$Cu$_4$Ni phases are partially dissolved in the α-Al matrix. As shown in Figure 3c, the sharp edges of the primary silicon become smooth. The size of block-like primary Si decreases with temperature increasing until 510 °C, and then, an increase in the primary Si size occurs at 520 °C. By increasing the solid solution treatment temperature from 480 to 500 °C (Figure 3b–d), fragmentation and spheroidization of the eutectic silicon particles are observed. With increased temperatures of 510 °C and 520 °C (Figure 3e,f), the silicon particles become significantly coarser, and the sizes of the silicon particle solid solution treated at 520 °C become larger than those at 510 °C. From the literature [39], it is observed that the fragmentation and spheroidization of eutectic silicon occurs first, after which the eutectic silicon becomes coarse. With solid
solution treatment temperature increasing, the time of eutectic silicon coarsening shortens. Therefore, whatever the temperature or time of the optimized solid solution treatment is, it should be chosen at the end of the first stage. The size of the eutectic silicon is obtained by measuring the maximum length of the acicular particles. As shown in Figure 3g, the size of the eutectic silicon decreases with increasing solid solution temperature until it reaches 510 °C and then increases over 520 °C. The size of the eutectic silicon of the Al-13Si-5Cu-2Ni alloy solid solution treated at 510 °C is 6.48 μm, which is 35.7% lower than that of the as-cast alloy (10.07 μm). Ostwald ripening appears on eutectic silicon when solid solution treatment is 520 °C, and then the size of eutectic silicon increases. Therefore, the optimal recommended solid solution time should be chosen at the end of the first stage. According to the size and distribution of the silicon phase, the optimal solution temperature can be determined to be 510 °C.

Figure 3. Microstructures of Al-13Si-5Cu-2Ni alloy solid solution treated at different temperatures for 8 h and 165 °C aging for 10 h and the size of eutectic silicon of alloys: (a) as cast, (b) 480 °C, (c) 490 °C, (d) 500 °C, (e) 510 °C, (f) 520 °C, and (g) the size of eutectic silicon.

Figure 4 shows the mechanical properties of the Al-13Si-5Cu-2Ni alloy solution treated at different temperatures for 8 h and 165 °C aging for 10 h. The corresponding tensile strength and fracture strain data are shown in Table 3. The tensile strength and fracture strain of the alloys increase simultaneously with increasing solid solution temperature in the temperature range of 480–510 °C. From Figure 4b, when the solid solution treatment temperature is 510 °C, the optimum tensile properties of the alloys are obtained with maximum values of 225 MPa (yield strength), 383 MPa (tensile strength), 2.8% (elongation), and 5.7% (fracture strain). The yield strength, tensile strength, elongation, and fracture strain of the alloy solution treated at 520 °C are 193 MPa, 332 MPa, 2.7%, and 5.6%, respectively, which are lower than those at 510 °C. Combined with Figure 3, the reason for this phenomenon is that the content of the Al₂Cu and Al₆Cu₄Ni phases on the dendrite grain boundaries dissolved in the Al matrix increases with the increase in solid solution treatment temperature. The dissolution of the Al₂Cu and Al₆Cu₄Ni phases reduces the fragmentation of the dendrite grain boundaries and the concentration of stress during alloy deformation. When the solid solution treatment temperature is lower than 510 °C, it is not sufficient to dissolve in the α-Al matrix for the Al₂Cu and Al₆Cu₄Ni phases located in
the dendrite grain boundaries. When the solid solution treatment temperature is higher than 510 °C, the eutectic silicon grows up because of Ostwald ripening, which reduces the mechanical properties of the alloys. Moreover, the silicon particles of the alloy treated at 520 °C grow and become coarse, resulting in a deterioration of the mechanical properties of the alloy. As a consequence, the optimal solid solution treatment temperature for Al-13Si-5Cu-2Ni alloy is 510 °C under a solid solution treatment for 8 h and aging treatment at 165 °C for 10 h Al-13Si-5Cu-2Ni alloy.

![Figure 4](image-url) The mechanical properties of Al-13Si-5Cu-2Ni alloy solution treated at different temperatures for 8 h and 165 °C aging for 10 h: (a) the engineering stress–strain curves and (b) bar graph of tensile strength and fracture strain.

| Sample | Temperature (°C) | Yield Strength (MPa) | Tensile Strength (MPa) | Elongation (%) | Fracture Strain (%) |
|--------|------------------|----------------------|------------------------|----------------|---------------------|
| a      | 480              | 203 ± 10             | 352 ± 11               | 2.9 ± 0.5      | 4.8 ± 0.4           |
| b      | 490              | 212 ± 9              | 365 ± 12               | 3.1 ± 0.4      | 5.3 ± 0.2           |
| c      | 500              | 223 ± 10             | 371 ± 7                | 3.0 ± 0.3      | 5.5 ± 0.3           |
| d      | 510              | 225 ± 11             | 383 ± 14               | 2.8 ± 0.1      | 5.7 ± 0.2           |
| e      | 520              | 193 ± 7              | 332 ± 14               | 2.7 ± 0.4      | 5.6 ± 0.3           |

### 3.3. Effect of Aging Treatment on the Microstructure and Tensile Properties of Al-13Si-5Cu-2Ni Alloy

The purpose of using T6 heat treatment for Al-13Si-5Cu-2Ni alloy is to dissolve the Al₂Cu phase and Al₇Cu₄Ni phase located at dendrite boundaries by solution heat treatment. Cu and Ni are supersaturated into the α-Al matrix using solid solution treatment, and then, the θ’ phase is uniformly precipitated. The aim of using an aging treatment is to strengthen precipitation and elevate the strength of alloys. There is a huge influence of aging treatment on the strength and hardness of alloys. If the aging treatment time is short, the θ’ phase is not sufficiently precipitated, and the improvement in the strength and hardness is not sufficient. If the aging treatment time is long, the θ’ phase will coarsen, and then the strength and hardness of the alloys will decrease. Therefore, it is important to optimize the appropriate aging treatment time. An aging temperature of 165 °C was selected to optimize the aging treatment time in this study.

Figure 5 shows the microstructures of the Al-13Si-5Cu-2Ni alloy solid solution treated at 510 °C for 8 h and aging treated at 165 °C for different hours. Figure 5a shows the metallurgy of the as-cast Al-13Si-5Cu-2Ni alloy microstructure. The dendrite grain boundaries are clearly observed in Figure 4a. Eutectic silicon is strip-shaped, and the edges of the primary silicon are sharp. Figure 5b–i shows the microstructure of the Al-13Si-5Cu-2Ni alloy solid solution treated at 510 °C for 8 h of aging and treated at 165 °C for different durations. Comparing Figure 5a,b, the edges of the block-like primary silicon change from being sharp to smooth. The amount and distribution of eutectic silicon becomes uniform. This is because the solid solution treatment influences the microstructure of primary silicon
and eutectic silicon. However, there is little difference between Figure 5b–i. It is difficult to affect primary silicon and eutectic silicon at an aging temperature of 165 °C. The size of the eutectic silicon decreases after solid solution treatment, but the change in eutectic silicon size is not obvious under different aging treatment times. TEM images were obtained to further understand the effect of aging time on the Al-13Si-5Cu-2Ni alloy microstructure.

As illustrated in Figure 6a–c, the $\theta'$ phase precipitated out after the solid solution at 510 °C for 8 h and aging at 165 °C for 6, 10, and 14 h in Al-Si$_{13}$-Cu$_{0.5}$-Ni$_2$ alloys. Figure 6 shows that the main precipitate phase is the $\theta'$ phase. As the aging time increases, the average diameter and thickness increase. The diameter statistics for Al-Si$_{13}$-Cu$_{0.5}$-Ni$_2$ alloys aged at 165 °C for different times are shown in Figure 6d,e. The volume fraction of the $\theta'$ precipitates is estimated to be 3.1%, 4.5%, and 4.7% for aging times of 6, 10, and 14 h. For an aging time of 6 h, the average diameter is 74.62 nm. For an aging time of 10 h, the average diameter is 82.13 nm. For an aging time of 14 h, the average diameter is 89.60 nm. The volume fraction and average diameter increase with the increasing aging time.

Figure 5. Microstructures of Al-13Si-5Cu-2Ni alloy 510 °C solid solution treated for 8 h and 165 °C aging for different hours and the size of eutectic silicon of alloys: (a) as cast, (b) 2 h, (c) 4 h, (d) 6 h, (e) 8 h, (f) 10 h, (g) 12 h, (h) 14 h, (i) and 16 h.

Figure 6. Precipitates of Al-Si$_{13}$-Cu$_{0.5}$-Ni$_2$ alloys after different hours of aging: (a) 6 h, (b) 10 h, and (c) 14 h. (d–f) The precipitates size distribution of these materials.
Figure 7a,b shows the mechanical properties of Al-13Si-5Cu-2Ni alloy solid solution treated at 510 °C for 8 h and aging treated at 165 °C for different durations. The corresponding normal temperature tensile strength and fracture strain data are shown in Table 4. In the range of 2–8 h, the tensile strength of alloys increases with increasing aging treatment time. When the solid solution treatment time is 8 h, the optimum tensile properties of the alloys are obtained with maximum values of 237 MPa (yield strength), 385 MPa (tensile strength), 4.5% (elongation), and 6.0% (fracture strain). In the range of 10–14 h, the tensile strength and fracture strain of alloys decrease with the increase in solid solution treatment time. Compared with other methods, including alloying, deformation, or different heat treatments, the mechanical properties of Al-13Si-5Cu-2Ni alloy in this work are found to be better, as shown in Figure 7c and Table 5. The hardness of the Al-13Si-5Cu-2Ni solution treated at 510 °C for 8 h and then aged at 165 °C for different hours is illustrated in Figure 7d. There is an evident trend of increasing hardness for the Al-13Si-5Cu-2Ni alloy with aging time increasing until 10 h, followed by a drop in the hardness. The micro-hardness of Al-13Si-5Cu-2Ni alloy aged for 10 h, 164 HV, is 18.84% higher than the hardness of unmodified Al-13Si-5Cu-2Ni alloy (138 HV). There is a huge influence of aging treatment on the strength and hardness of the alloys. If the aging treatment is short, the $\theta^\prime$ phase is not sufficiently precipitated, and the improvement in the strength and hardness is not sufficient. If the aging treatment is long, the $\theta^\prime$ phase will coarsen, and then the strength and hardness of alloys will decrease. Not much difference was found when comparing the hardness of samples subject to an aging treatment for 8 and 10 h. According to a comparison of the tensile strength, fracture strain, and hardness of Al-13Si-5Cu-2Ni alloy under different T6 heat treatments, the best mechanical properties of this alloy are obtained after solid solution treatment at 510 °C for 8 h and aging treatment at 165 °C for 8 h.

**Figure 7.** The mechanical properties and hardness of Al-13Si-5Cu-2Ni alloy solid solution treated at 510 °C for 8 h and 165 °C aging for different hours: (a) the engineering stress–strain curves, (b) bar graph of tensile strength and fracture strain, (c) hardness of aging different hours, and (d) contrast diagram of different strengthening methods for Al alloys with this work.
4. Discussion

The solid solution treatment promotes the solid solution of the elements into the matrix and provides the driving force for the fine and dispersed second phase precipitations during aging treatment. As shown in Figure 8, there exist Al$_2$Cu phases and Al$_3$Cu$_4$Ni phases at the grain boundaries of the as-cast Al-Si$_{13}$-Cu$_{0.5}$-Ni$_2$ alloys. In addition, the solid solution treatment also leads to passivation of primary silicon and fragmentation and spheroidization of eutectic silicon. This reduces the split between silicon phases and the matrix in the Al-Si alloys. Therefore, the size of eutectic silicon is the smallest when solid solution treatment temperature is 510 °C, and then the tensile strength of sample solution treated at 510 °C is the highest. The θ' phases precipitated during aging treatment pin the dislocations under load and then improve the strength of the alloys, which is called Orowan strengthening. We used the modified Orowan equation to calculate the Orowan stress due to plate-like θ' precipitate strengthening ($\Delta \sigma_{\theta'}$) according to the following modified Orowan equation [39,46–48]:

$$\Delta \sigma_{\theta'} = \frac{M G b}{2 \pi \sqrt{1 - \nu}} \left( \frac{1}{1.123 d} \sqrt{\frac{0.318 a f}{a} - \frac{\pi d}{8} - \frac{0.061 d}{a}} \right) \ln \frac{0.981 d}{ab}, \quad (1)$$

where $M$ is the Taylor factor for Al (approximately equal to 3 [39]), $G$ is the shear modulus (26 GPa), $\nu$ is the Poisson’s ratio (1/3 for Al [39]), $b$ is the Burgers vector (0.286 nm for Al [39]), $a$ is the aspect ratio of the precipitates ($a = d/t$, where $d$ and $t$ are the diameter and thickness of the precipitates, respectively), and $f$ is the volume fraction of the θ' precipitates. It is known that a smaller diameter and greater number density for the θ' precipitated phases can produce a more significant strengthening effect to improve the yield strength. $\Delta \sigma_{\theta'}$ is 70, 83, and 75 MPa when aging time is 6, 10, 14 h through Equation (1), respectively. With increasing aging treatment time, the diameter of the θ' phase increases. A suitable aging treatment is extremely important.
5. Conclusions

The present work investigated the influence of heat treatment temperature and time on the microstructure and mechanical properties of Al-Si_{13}-Cu_{0.5}-Ni_{2} alloys. The optimization of T6 heat treatment parameters for Al-13Si-5Cu-2Ni alloy was investigated. The following conclusions can be drawn:

1. After solid solution treatment, the Al_{2}Cu phase and a part of the Al_{2}Cu_{4}Ni phase at the dendrite boundary are solid-dissolved into the α-Al matrix. As the solid solution treatment temperature increases until 510 °C, the sharp corners of the primary silicon are passivated, and the strip-shaped eutectic silicon melts and spheroidizes. Beyond 520 °C, a coarsening of the eutectic silicon occurs. The optimum tensile strength and fracture strain for an Al-13Si-5Cu-2Ni alloy solid solution treated at 510 °C is 383 MPa and 5.7%, respectively.

2. Different aging treatment times affect the amount and size of the θ′ precipitation phase, which then affects the properties of the materials. As the aging treatment time increases, the volume fraction, diameter, and length of the θ′ phase increases. The optimized T6 heat treatment involves solid solution treatment at 510 °C for 8 h followed by aging at 165 °C for 8 h, and the tensile strength and fracture strain of Al-13Si-5Cu-2Ni alloy processed to this heat treatment is 385 MPa and 6.0%, respectively.

3. After T6 heat treatment, the size and amount of silicon phases and θ′ phases determine the performance of the Al-13Si-5Cu-2Ni alloy. The size of the eutectic silicon is the smallest when the solution treatment is 8 h, which effectively reduces stress concentration, wakens crack propagation, and elevates the performance of the alloy. A smaller diameter and greater number density for the θ′ precipitated phases can produce a more significant Orowan strengthening. The solid solution treatment and aging treatment improve the mechanical properties of the Al-13Si-5Cu-2Ni alloy through passivation and spheroidization of the silicon phase and Orowan strengthening.

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