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# Effect of Heat Treatment on the Microstructure and Mechanical Properties of a Composite Made of Al-Si-Cu-Mg Aluminum Alloy Reinforced with SiC Particles

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**Abstract:** In this paper, the effect of heat treatment (solution treatment and artificial aging) on the microstructure and properties of as-cast Al5Si1Cu0.5Mg aluminum alloy and its composite reinforced with 1.5 wt.% SiC particles was studied. The results showed that at 520 °C the optimal solution time for the aluminum alloy and its composite is 9 h and 6 h, respectively. After solution treatment, the microstructure of these two materials consists of a uniform distribution of nearly spherical eutectic Si and skeletal $\gamma$ phase, furthermore, the composite eutectic Si phase is smaller and $\gamma$ phase is more dispersed. After artificial aging at 175 °C for 6 h, the microstructure of the composite is more dispersed and finer than that of the aluminum alloy on the whole and $Al_2Cu$ is precipitated. After heat treatment, the microhardness, ultimate tensile strength, and elongation of the aluminum alloy and its composite are higher than those of the as-casts. At the same time, the morphology of tensile fracture surface changes very much from a large area of cleavage plane to a large number of dimples and the tearing ridges become thicker for both the aluminum alloy and its composite.

**Keywords:** Al5Si1Cu0.5Mg aluminum alloy; SiC reinforced composite; heat treatment; microstructure; mechanical properties

## 1. Introduction

Aluminum alloys are primary materials for potential applications such as aerospace precision instruments because of their excellent properties, which include low density, high specific strength, and good hot workability. In particular, the castings are preferred in aircraft manufacturing industries, and their hardness, strength, and elongation can be improved greatly by heat treatment [1,2]. In order to further improve mechanical properties of aluminum alloys, the particles of SiC or $Al_2O_3$ (SiC and $Al_2O_3$ are abbreviations of silicon carbide and aluminum oxide, respectively) may be added to develop aluminum matrix composites [3,4]. Al5Si1Cu0.5Mg belongs to Al-Si-Cu-Mg series cast alloys with good casting performance and little linear shrinkage and can be strengthened through heat treatment. Now, Al5Si1Cu0.5Mg aluminum alloy is widely used in aircraft engines and airframes to mainly manufacture thin-walled, complex, and high load-bearing casting.

T6 heat treatment is usually employed to improve comprehensive mechanical properties of materials and consists of [5–9]: (i) solution heat treatment of as-cast samples for dissolution of certain intermetallic phases, such as $Al_2Cu$, some phases containing copper and magnesium elements, and for changing the morphology of eutectic silicon phase; (ii) quenching, usually at room temperature, to obtain a supersaturated solid solution; (iii) age hardening, namely, the precipitation of different types of precipitated phases such as $Mg_2Si$, $Al_2Cu$ and $Al_5Mg_8Si_6Cu_2$ from the supersaturated solid solution either at room temperature (natural aging) or at an elevated temperature (artificial aging). During T6 heat treatment, many researchers [7,10,11] found that solution time, aging time, aging temperature, and some other parameters have a great influence on the microstructure and properties of aluminum alloys. For example, Costa et al. [7] found that for a directionally solidified Al-Si-Cu alloy, the profile of secondary dendrite arm spacing values obtained along the length of the DS casing increased only when the solution time adopted in the T6 heat treatment was increased from 5 to 8 h. They also found that the samples heat treated with a solution time of 8 h exhibited a decrease in hardness when compared to that of samples subjected to a solution time of 5 h, which may be associated with the corresponding increase in the dendritic arm spacing, coarsening of Si particles, and collapse of tertiary dendritic arms that makes the dendritic network less complex. Ceschini et al. [11] found that the hardness, yield strength, and ultimate tensile strength of as-cast CAl5Si1Cu0.5Mg alloy, the chemical composition of which is similar to that of Al5Si1Cu0.5Mg alloy, were obviously reduced after overaging treatment, which was mainly caused by the coarsening of precipitated phases in the crystal. Thus, it is of importance to select the parameters of heat treatment to improve microstructure and mechanical properties of aluminum alloys.

The aim of this paper was to evaluate the effect of T6 heat treatment on the microstructure and mechanical properties for Al5Si1Cu0.5Mg aluminum alloy and its composite reinforced with 1.5 wt.% SiC particles. At the same time, the influence of the addition of SiC particles on the microstructure and mechanical properties of Al5Si1Cu0.5Mg aluminum alloy was also studied before and after T6 heat treatment.

## 2. Experimental Procedure

The matrix alloy used for this study is Al5Si1Cu0.5Mg alloy, which has the following chemical composition in weight percentage: 5.13 Si, 1.27 Cu, 0.55 Mg, 0.16 Ti, 0.13 Fe, and the balance Al. As the reinforcements, β-SiC particles in the weight percentage of about 1.5% with a diameter between 1 and 3 μm were added to Al5Si1Cu0.5Mg alloy to develop a composite. The solution and aging treatment for the matrix alloy and composite were conducted by using a KSG-24-16C resistance furnace according to the specifications shown in Table 1. During solution treatment, the samples were put into a resistance furnace after the temperature was stable at 520 ± 2 °C. For the matrix alloy, the soaking time was 1, 6, 9, and 12 h, respectively. For composite, the soaking time was 1, 3, 6 and 10 h, respectively. After solution treatment, the samples were quickly taken out and put into room temperature water for quenching. Then, all the samples were held at 175 ± 2 °C for 6 h for aging treatment and taken out into the air.

**Table 1.** Heat treatment parameters employed for Al5Si1Cu0.5Mg matrix alloy and composite.

| Materials | Solution Treatment (520 ± 2 °C) | Aging Treatment (175 ± 2 °C) |
|---|---|---|
| Al5Si1Cu0.5Mg | 1, 6, 9, 12 h | 6 h |
| Composite | 1, 3, 6, 9 h | 6 h |

Different samples of the as-cast and T6 heat treatment were ground using metallographic sandpaper from coarser to finer ranges. Subsequently, cloth polishing was applied using fine $Al_2O_3$ powder. The samples were then polished with Silvo solution, consisting of iso-propyl alcohol, ammonium hydroxide, and $SiO_2$ powder, which was used for cleaning and suspending the tarnish on surfaces and also to protect from oxidation [12]. Finally, the samples were etched by an acid solution of 0.5% HF and 99.5% $H_2O$. The microstructures of these samples were observed via scanning electron microscopy and light microscopy approaches. Chemical composition analysis was carried out by using an energy dispersive spectrometer equipped with scanning electron microscopy. Microhardness tests were conducted on the as-cast and T6 heat treated samples by a HVS-1000A Vivtorinox hardness tester using a 300 g load and a dwell time of 10 s. The adopted microhardness values are the average of at least 7 different measurements of each sample. In order to obtain the tensile properties, the sample was cut into a rod with a diameter of 6 mm and a gauge of 30 mm as shown in Figure 1. The tensile test was conducted on a SUNS UTM5105 electronic universal tensile testing machine with a rate of 1.0 mm/min. The fractography was observed by using a scanning electron microscope.

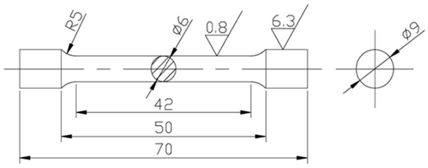


**Figure 1.** Dimensions of the tensile specimen.

## 3. Results and Discussion

### *3.1. Solution Treatment*

At the solution temperature of 520 °C for different soaking times, the microstructures of Al5Si1Cu0.5Mg matrix alloy and its composite are shown in Figure 2. Generally speaking, the microstructure of as-cast Al5Si1Cu0.5Mg alloy is mainly composed of the coarse primary $\alpha$-Al dendrites, the long acicular eutectic Si phase, and the stacked skeletal iron-rich $\gamma$ phase. During the solution treatment, the elements of Si, Cu, Mg, and others would be dissolved and diffused to a certain degree at high temperature with increasing soaking time. It can be seen from Figure 2a that for Al5Si1Cu0.5Mg matrix alloy after being held at 520 °C for 1 h, the eutectic Si phase exhibited a long needle-like shape, while the iron-rich $\gamma$ phase is stacked and skeletal. This indicated that the soaking time for obtaining the optimal solution microstructure was not enough. When the soaking time was increased to 6 h, the morphology of the eutectic Si phase changed obviously, as shown in Figure 2c. Most of the thick, long needle-like eutectic Si phases had been changed into near sphericity, and a small amount of them are like fine needles. At the same time, the morphology of the iron-rich $\gamma$ phase was changed significantly. Most of the stacked and skeletal iron-rich $\gamma$ phase became scattered and there was a certain spacing among skeletons. Because inhomogeneous microstructure leads to the degeneration of mechanical properties, it is considered that increasing the soaking time would be helpful to the diffusion of alloy elements and would improve the uniformity of the microstructures. When the soaking time was increased to 9 h, all the needle-like eutectic Si phases were transformed into near sphericity and distributed evenly, and no stacked $\gamma$ phase existed, as shown in Figure 2e. When the soaking time was increased to 12 h, it can be seen from Figure 2g that the eutectic Si phase and $\gamma$ phase in the alloy were coarsened, namely by Ostwald ripening [13]. For composite, the evolution of the microstructure with the soaking time is similar to that of the Al5Si1Cu0.5Mg matrix alloy. That is to say, when the soaking time reached about 6 h, all the needle-like eutectic Si phases were transformed into near sphericity and distributed evenly and, at the same time, no stacked $\gamma$ phase existed.

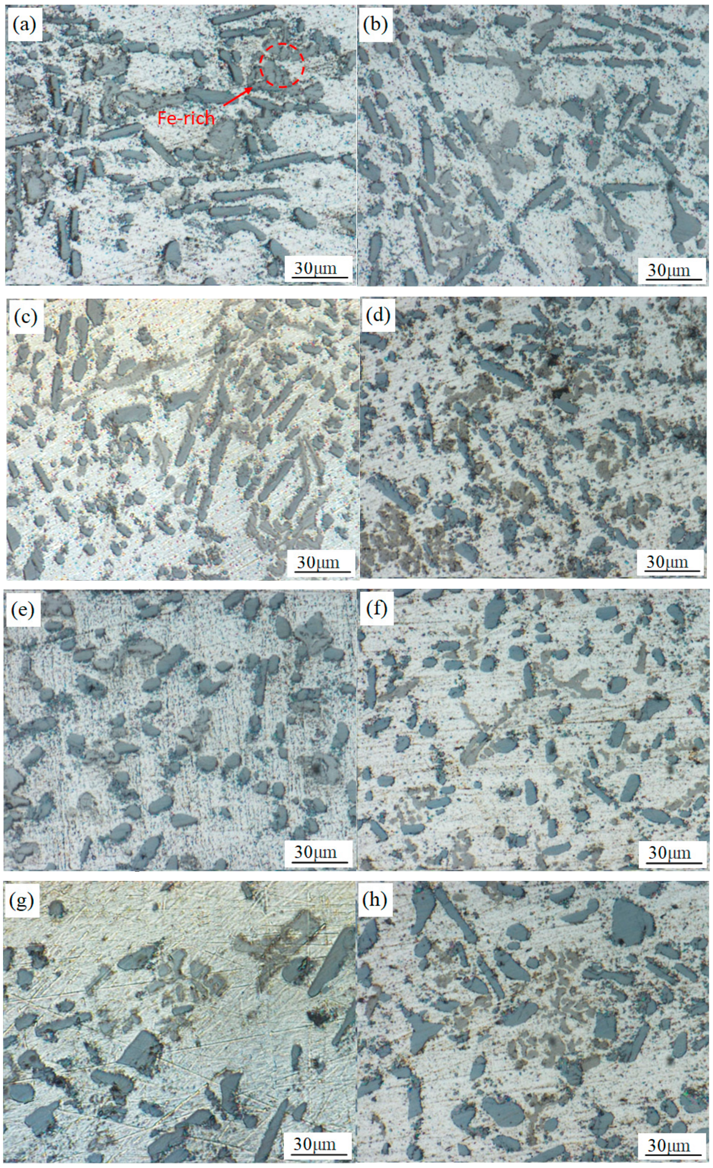


**Figure 2.** Microstructures of Al5Si1Cu0.5Mg matrix alloy and composite at 520 °C for different soaking time: (**a**) 1 h, (**c**) 6 h, (**e**) 9 h, and (**g**) 12 h for matrix alloy, and (**b**) 1 h, (**d**) 3 h, (**f**) 6 h, and (**h**) 10 h for composite.

The backscattered electron images for as-cast and solution-treated composites for 6 h are shown in Figure 3. Figure 4 is the EDS patterns of areas of A and B marked in Figure 3b, which indicated that the bright short rod or block in the as-cast composite was the $Al_2Cu$ phase. In order to accurately identify the evolution of the acicular eutectic Si phase and whether the $Al_2Cu$ phase was completely dissolved after solution treatment with 6 h, the region marked in Figure 3c was enlarged and shown in Figure 3d. It can be seen from Figure 3d that the acicular eutectic Si phase had been mostly spherical and there existed obvious spacing among the skeleton-like $\gamma$ phases. In addition, the bright short rod or block $Al_2Cu$ phase cannot be found. This indicated that the $Al_2Cu$ phase had been almost dissolved. Considering the nearly spherical eutectic Si phase with its small size and dispersed distribution as well as apparent coarsening of the microstructures after the solution treatment with 10 h, it can be inferred that the optimal time for solution treatment is about 6 h.

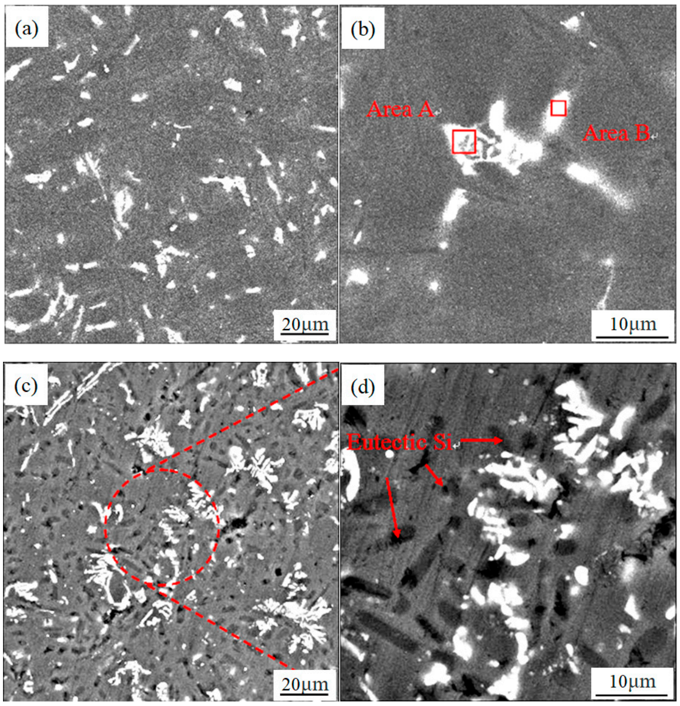


**Figure 3.** Backscattered electron images of (**a**,**b**) as-cast and (**c**,**d**) solution-treated (6 h) composites.

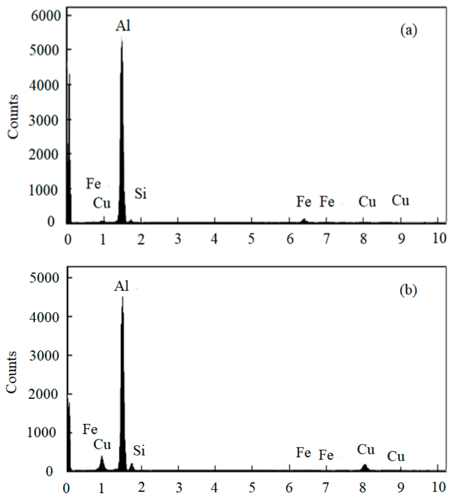


**Figure 4.** EDS analysis of areas of (**a**) A and (**b**) B marked in Figure 3b.

### 3.2. Aging Treatment

After the soaking time of 6 h at the aging temperature of 175 °C, the microstructures of Al5Si1Cu0.5Mg matrix alloy and its composite were shown in Figure 5a,b, respectively. From Figure 5 it can be seen that the microstructure of composite with SiC particles was more dispersed and fine than that of Al5Si1Cu0.5Mg matrix alloy, which was helpful to improve the mechanical properties. In addition, the secondary precipitation phases in Al5Si1Cu0.5Mg matrix alloy and its composite cannot be obviously observed which may be due to their small size, therefore limiting the use of an optical microscope [14]. The SEM images of the composite after aging treatment were shown in Figure 6. Figure 6b was the enlargement of a local area marked by a red circle in Figure 6a. In Figure 6b, the skeletal iron-rich γ phase with cleat outline and nearly spherical eutectic Si phase were uniformly distributed, but $Al_2Cu$, $Mg_2Si$, and other precipitates were not observed.

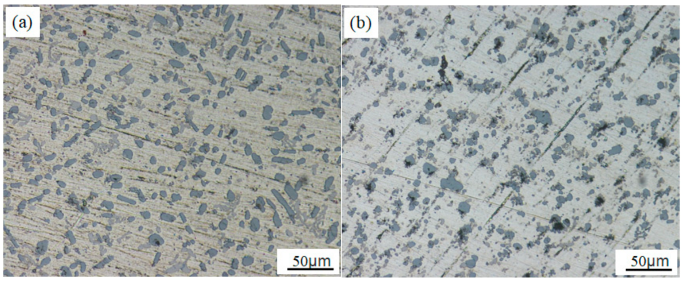


**Figure 5.** Microstructures of (**a**) Al5Si1Cu0.5Mg matrix alloy, and (**b**) composite after aging treatment.

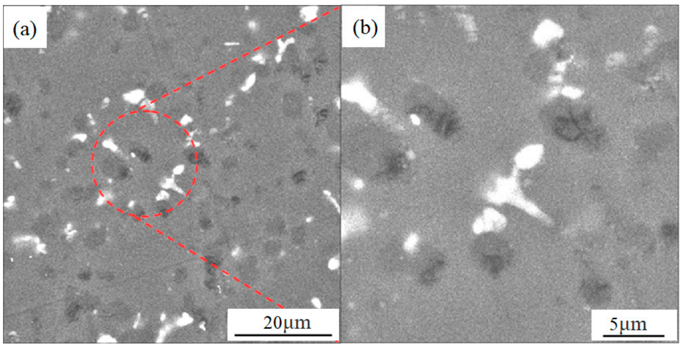


**Figure 6.** SEM images of the composite after aging treatment: (**b**) is an enlargement of a local area marked in (**a**).

Wang et al. [15] reported that for as-cast A356, precipitation hardening occurred in Al matrix which was enriched with Mg and Si elements after T6 heat treatment. Zhu et al. [16] also reported that the long rod of the precipitated phase $Mg_2Si$ was dispersed in an $\alpha$-Al matrix for the same alloy. For Al-Si-Cu-Mg alloy, the precipitated phases such as $Mg_2Si$, $Al_2Cu$, and $Al_5Mg_8Si_6Cu_2$ would be dispersed after an aging treatment [9]. In another article that the corresponding author published in 2018, the fine needle-like $Al_2Cu$ precipitates with about 100 nm length can be observed with high density dislocations in the same composite [17].

For the composite, the suitable aging time was shorter than that of Al5Si1Cu0.5Mg matrix alloy, and the reason is that the coefficient of thermal expansion of SiC is different from Al matrix alloy which resulted in the existence of high-density dislocations around the SiC particles. This would increase the rate of diffusion of solid solution elements and reduce the nucleation activation energy of the precipitated phase [18]. At the same time, the SiC particles can pin grain boundaries and stabilize the cellular substructure, which will accelerate the aging reaction and increase the rate of work hardening [19]. However, if the soaking time for aging treatment was too long, overaging would occur and degenerate the microstructures and properties [14].

### *3.3. Effect of T6 Heat Treatment on Mechanical Properties of Al5Si1Cu0.5Mg Matrix Alloy and Composite*

#### 3.3.1. Microhardness

Figure 7 showed the Vickers-hardness (HV) values of Al5Si1Cu0.5Mg matrix alloy and its composite before and after T6 heat treatment. It can be obviously seen from Figure 7 that the Vickers-hardness values of the composite and Al5Si1Cu0.5Mg matrix alloy were 87.06 HV and 80.17 HV before T6 heat treatment and were 101.86 HV and 91.81 HV after T6 heat treatment, respectively. The microhardness of the composite was about 8.59% and 10.95% higher in percentage than that of Al5Si1Cu0.5Mg matrix alloy before and after T6 heat treatment. The percentage increase in microhardness values was 14.52% for Al5Si1Cu0.5Mg matrix alloy and 17.00% for the composite after T6 heat treatment.

Thus, two important observations can be made. Firstly, the microhardness of the composite, irrespective of T6 heat treatment, was consistently higher than that of Al5Si1Cu0.5Mg matrix alloy.

This was due to the strengthening effect of the reinforced SiC particle itself and the hardening effect of the high density dislocations induced by the difference of the thermal expansion coefficient between the SiC particle and the matrix [20] or uniform distribution of reinforcement and resistance to deformation from the ceramic phases [4]. Secondly, T6 heat treatment is helpful to increase the microhardness of both Al5Si1Cu0.5Mg matrix alloy and its composite. The reason is as follows: At the stage of solution treatment, the eutectic Si phase became necked, separated, and spheroidized and would cause fine grain strengthening. At the same time, the secondary precipitation phases, such as $Al_2Cu$, were dissolved to homogenize the alloy composition. Then, in the stage of aging treatment, the hard phase was dispersively precipitated, which is also important in enhancing the hardness of the material [9,16].

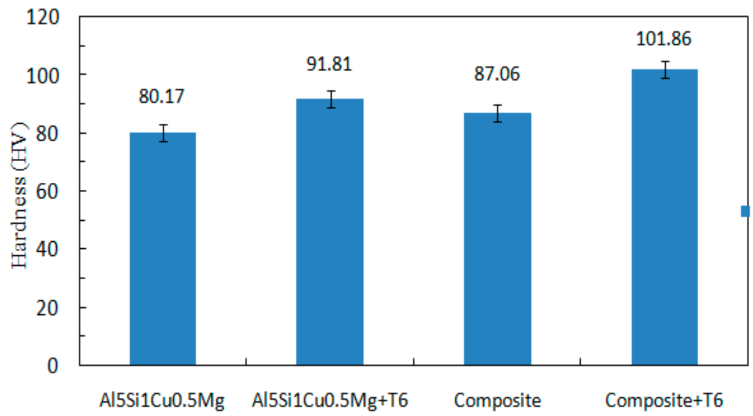


**Figure 7.** Effect of T6 heat treatment on microhardness of Al5Si1Cu0.5Mg matrix alloy and composite.

### 3.3.2. Tensile Properties

Table 2 showed the values of ultimate tensile strength (UTS) and elongation for Al5Si1Cu0.5Mg matrix alloy and its composite before and after T6 heat treatment, respectively. By analyzing the data in Table 2, two important discoveries can be made. Firstly, the values of ultimate tensile strength and elongation of the composite, irrespective of T6 heat treatment, were consistently higher than those of Al5Si1Cu0.5Mg matrix alloy. That is to say, the addition of 1.5 wt.% SiC particles can improve ultimate tensile strength and elongation for Al5Si1Cu0.5Mg aluminum alloy. Secondly, after T6 heat treatment, the values of ultimate tensile strength or elongation of Al5Si1Cu0.5Mg matrix alloy and its composite were consistently higher than those of the as-casts. This indicated that T6 heat treatment plays an important role in improving ultimate tensile strength and elongation. The increase in ultimate tensile strength and elongation may be attributed to the following [2,4,21]: (i) SiC particles added into Al5Si1Cu0.5Mg matrix alloy act as nucleation sites and lead to a structure with small grain size, which helps to improve not only ultimate tensile strength but also elongation; (ii) the difference in the thermal expansion coefficient between the matrix alloy and SiC particle can induce high-density dislocations, and the fine $Al_2Cu$ can be precipitated after T6 heat treatment which helps to improve the ultimate tensile strength; (iii) T6 heat treatment leads to the morphological transition for eutectic Si from a coarse acicular shape to near sphericity and for the iron-rich $\gamma$ phase from stack to scatter. This will reduce the cutting effects of the acicular eutectic Si, stacked, and skeletal iron-rich $\gamma$ phase on the matrix alloy or composite. In addition, T6 heat treatment would be helpful to the diffusion of alloy elements and for improving the uniformity of the microstructures. All the above are favorable in improving the ultimate tensile strength and elongation.

**Table 2.** Effect of T6 heat treatment on ultimate tensile strength (UTS) and elongation of Al5Si1Cu0.5Mg matrix alloy and composite.

| Parameters \ Samples | Al5Si1Cu0.5Mg | Composite | Al5Si1Cu0.5Mg + T6 | Composite + T6 |
|---|---|---|---|---|
| UTS/MPa | 191.56 ± 2.00 | 238.62 ± 2.00 | 246.54 ± 2.00 | 273.64 ± 2.00 |
| Elongation/% | 3.81 ± 0.50 | 5.23 ± 0.50 | 5.94 ± 0.50 | 6.12 ± 0.50 |

### 3.3.3. Fractography

The SEM images in Figure 8 show the tensile fracture surface of Al5Si1Cu0.5Mg matrix alloy and its composite before and after T6 heat treatment. In Figure 8a, a large area of cleavage plane and some tearing ridges can be observed and there are almost no dimples, which is the characteristic of a cleavage fracture. Considering that the elongation is only 3.81%, the fracture of as-cast Al5Si1Cu0.5Mg matrix alloy is brittle. In the microstructure of as-cast composite shown in Figure 8b, there are some dimples and tearing ridges, but the area of smooth cleavage plane is decreased. Furthermore, the tearing ridges are thicker than that of as-cast Al5Si1Cu0.5Mg matrix alloy. This indicates that the as-cast composite had a certain degree of plastic deformation and the elongation is about 5.23% with an increase of 37.17% compared with that of Al5Si1Cu0.5Mg matrix alloy. After T6 heat treatment, the morphology of the fracture surface changed greatly for both Al5Si1Cu0.5Mg matrix alloy and its composite. From Figure 8c, it can be seen that there exist many dimples with different sizes and some thicker tearing ridges, but no smooth cleavage plane. This suggests that after T6 heat treatment, the plasticity of Al5Si1Cu0.5Mg matrix alloy can be improved. The elongation increases to 5.94%. The fracture morphology of the composite after T6 heat treatment is given in Figure 8d. It can be seen from Figure 8d that there are a large number of dimples. Compared with those in Figure 8c (Al5Si1Cu0.5Mg matrix alloy after T6 heat treatment), the dimples are deeper and distributed more evenly, thus the plasticity of the composite is improved and the elongation increases from 5.94% to 6.12%.

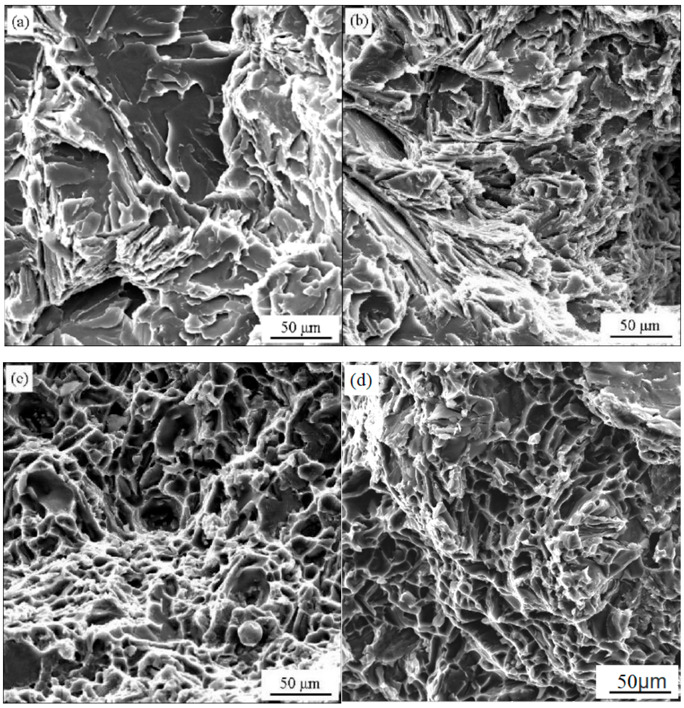


**Figure 8.** SEM images of tensile fracture surface: (**a**) before and (**c**) after T6 for Al5Si1Cu0.5Mg matrix alloy, and (**b**) before and (**d**) after T6 for composite.

## 4. Conclusions

1. The solution treatment of Al5Si1Cu0.5Mg matrix alloy and its composite leads to the morphological changes of the eutectic Si and iron-rich $\gamma$ phase and causes the $Al_2Cu$ phase to dissolve. At 520 °C, the optimal time of solution treatment is 9 h and 6 h for Al5Si1Cu0.5Mg matrix alloy and its composite, respectively.
2. After an aging treatment with 6 h at the temperature of 175 °C, the microstructure of the composite is more dispersed and finer than that of Al5Si1Cu0.5Mg matrix alloy on the whole and $Al_2Cu$ will be precipitated.
3. Irrespective of T6 heat treatment, the microhardness, ultimate tensile strength, and elongation of the composite are consistently higher than those of Al5Si1Cu0.5Mg matrix alloy. The microhardness, ultimate tensile strength, and elongation of Al5Si1Cu0.5Mg matrix alloy and its composite after T6 heat treatment are higher than those of the as-casts.
4. After T6 heat treatment, a large area of cleavage plane is replaced by a large number of dimples and the tearing ridges become thicker in the fracture surfaces of both Al5Si1Cu0.5Mg matrix alloy and its composite.

**Author Contributions:** X.L. and H.Y. conceived and designed the experiments; X.L. and Z.-W.W. performed the experiments; X.L., H.Y., J.-L.L. and Q.N. analyzed the data; X.L., H.Y. and Z.-W.W. contributed reagents/materials/analysis tools; X.L., H.Y., Z.-W.W., N.L., J.-L.L. and Q.N. wrote the paper.

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