Article ID: 1003－6326(2004)03－0496－05

Effect of heat treatment on eutectic silicon morphology and mechanical property of Al-Si-Cu-Mg cast alloys①

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Abstract: Through morphology observation on silicon particles of Al-Si-Cu-Mg cast alloy, it is found that during solution treatment the evolution of eutectic silicon morphology and their effect on mechanical properties can be classified into three stages. In the initial stage, necking, stubbing and fragmentation of silicon particles result in the improvement of plasticity of alloy. In the intermediate stage, the mechanical properties of 354 alloy attain peak values due to spheroidization of silicon particles. In the final stage, the drop of hardness and strength is related to the deterioration of silicon morphology. The facets and lap occur in silicon particles and the coarsening process of silicon follows LSW model. During aging, the clusters of excess silicon can work as barriers for dislocation movement and thus enhance the strength of alloy. On the other hand, excess Si affects the process of aging precipitation and leads to a fine and highly dense distribution of GP zones, finally effectively strengthens the alloy.

Key words: Al-Si-Cu-Mg alloy; eutectic silicon; mechanical property; heat treatment; quantitative metallography

CLC number: TG 136; TG146.2 **Document code:** A

1 INTRODUCTION

Cast Al-Si base alloys are one of the most important aluminum alloys attributed to their low density, high specific strength, excellent casting properties, mechanical properties and erosion resistance. These Al-Si alloys are mainly applied in automotive, aerospace, transportation and mechanical industry, etc[1, 2]. The properties of alloys can be improved through adding Cu and Mg elements into the alloys. In the process of solution treatment, alloying elements completely dissolve and form supersaturated solid solution. However, due to high temperature diffusion, the morphology of eutectic silicon changes largely, resulting in a significant influence on the mechanical properties of Al-Si-Cu-Mg alloy.

In the past decade, many researchers reported the evolution of silicon phase in Al-Si alloys during solution treatment[3－6], and some works have been undertaken to address the excess Si in Al-Si base cast alloys[7－10]. Zhang[7] has reported that during ageing, the excess amount of silicon, which is not combined with magnesium to form β″ phase precipitates, may remain in solid solution, or precipitate in the form of pure silicon particles, or diffuse into existing eutectic silicon particles. Existence of a large number of silicon precipitates in the aged microstructure suggests that the excess silicon can form silicon precipitates easier than diffuse to existing eutectic silicon particles, although in the mechanism of the latter process it does not need to overcome any nucleation barrier.

There are, however, little published data describing the evolution of silicon morphology in the whole process of solution and aging treatments, especially about Al-Si-Cu-Mg quaternary alloy. Therefore, in the present study, the evolution of silicon morphology during solution treatment, the precipitation of excess silicon during age treatment and their influences on mechanical properties are investigated on Al-Si-Cu-Mg alloy by quantitative metallographic measurement, SEM and TEM observation and mechanical properties measurement.

2 EXPERIMENTAL

The chemical composition of tested alloy is listed in Table 1. The alloy was remelted at 750 ℃ and modified using ternary sodium-salts modifier (1.0%, mass fraction). Test bars (gauge length of 5 mm and

① **Received date:** 2003－09－02; **Accepted date:** 2003－12－15

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cross-section diameter of 12 mm) were solution treated at 525 ℃ for different time in a high-temperature intelligent numerical controlled furnace (where the temperature could be controlled within ±5 ℃) under the protection of nitrogen gas, then quenched into hot water (60 - 80 ℃), and subsequently aged at 175 ℃ for different times in drying oven(where the temperature could be controlled within ±2 ℃). The aging delay is less than 10 s. For each individual treatment at least 6 test bars were used.

Table 1 Chemical composition of alloy (mass fraction, %)

Si	Cu	Mg	Mn	Ti	Al
9.50	1.48	0.43	0.22	0.23	Bal.

Hardness measurement was preformed on a Brinell hardness tester with a load of 2.5 kN and a dwell time of 30 s. The Brinell hardness value (HB) was obtained by taking average values of at least 6 measurements for each state. Tensile test bars with gauge length of 25 mm and diameter of 5 mm were made according to GB145—59. Ultimate tensile strength(UTS) and elongation of alloys were obtained by using WE-10A tensile-tester. TEM samples were mechanically polished to 70 nm followed by twin electrolytic jet polishing using 30% nitric acid and 70% methanol etchant(volume fraction) maintained at - 20 ℃. Microstructure of samples were observed by TEM (PHILIPS EM 40) operated at 100 kV.

To obtain a more detail impression of the silicon morphology within the eutectic areas, the aluminum matrix was removed by etching. SEM etchant consisted of 12 mL HCL, 10 mL HF and 90 mL distilled water. The samples were rinsed in cold water and dried well before examination in S-570 scanning electron microscope. The evolution of eutectic silicon morphology during solution treatment was measured by software ISA4 and described by using various parameters of silicon particles such as mean area, area equivalent diameter, area ratio, length/width ratio, silicon particles counts per unit area etc.

3 RESULTS AND DISCUSSION

3.1 Effect of solution treatment on mechanical properties

Fig. 1 shows the effect of solution treatment on hardness, UTS and elongation of the alloy(T4 and T6 treatment). With increase of solution time up to 10 - 12 h, the hardness, UTS and elongation increase. After peak value the mechanical properties decrease, especially hardness and UTS. Fig. 2 shows SEM fractographs of as-cast alloy and solution treated alloys at 525 ℃ for 10 h and 14 h respectively. The fracture surface of quenched samples with rocky pattern, however, transforms to equiaxed dimple after being solution treated for 10 h, which results from the improvement of the ductility. Further solution leads to larger equiaxed dimples on fracture surface (Fig. 2(c)).

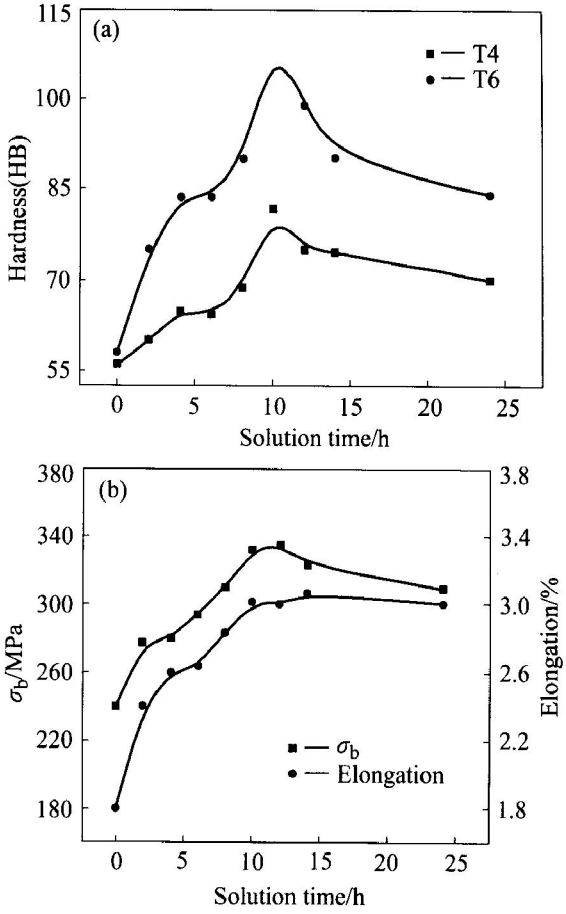

Fig. 1 Mechanical properties of 354 alloy as function of solution time at 525 ℃
(a) —Hardness (T4 and T6 treatment);
(b) — σ_b and Elongation (T6 treatment)

3.2 Effect of solution treatment on silicon morphology

Fig. 3 shows the change of eutectic silicon morphology during solution treatment. It can be found that the morphology of silicon changes remarkably. In the as-cast samples, eutectic fibrous-like eutectic silicon can be observed (Fig. 3(a)). Figs. 3(b), (c) and (d) show the process of eutectic silicon morphology evolution in the early stage of solution treatment (2 - 10 h). Silicon particles undergo necking and fragmentation in special positions. Afterwards the fragmentation and stubbing of silicon phase finish. The growth of silicon particles becomes slowly and the silicon particles gradually become rounder (Fig. 3(d) and 3(e)). When prolonging the solution time to 10 h, the fine silicon particles disappear, and the large ones grow up obviously. The morphology of silicon occurs in facets and lap, and the roundness of silicon particles does not increase continuously (Fig. 3(f)).

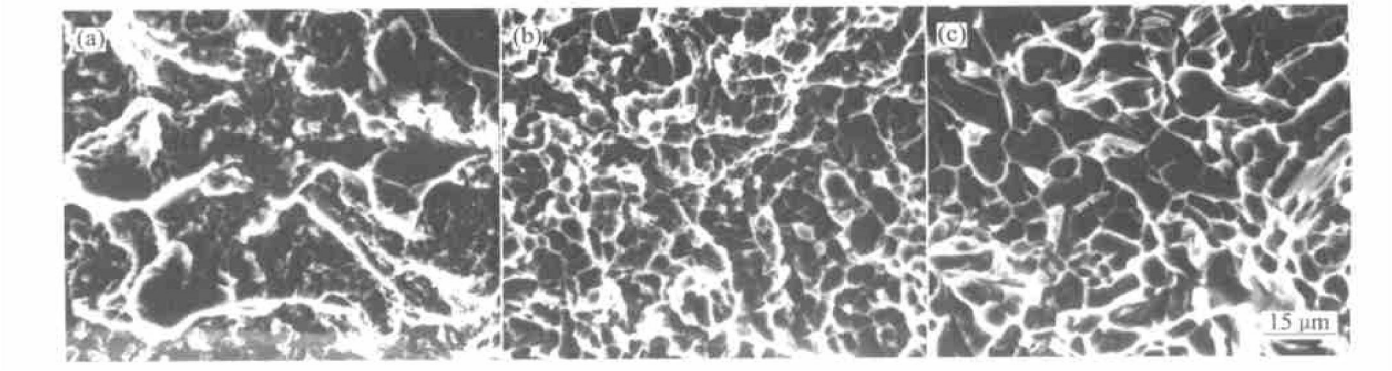

Fig. 2 SEM fractographs of alloys
(a) —As-cast; (b) —Solution treated at 525 ℃ for 10 h; (c) —Solution treated at 525 ℃ for 14 h

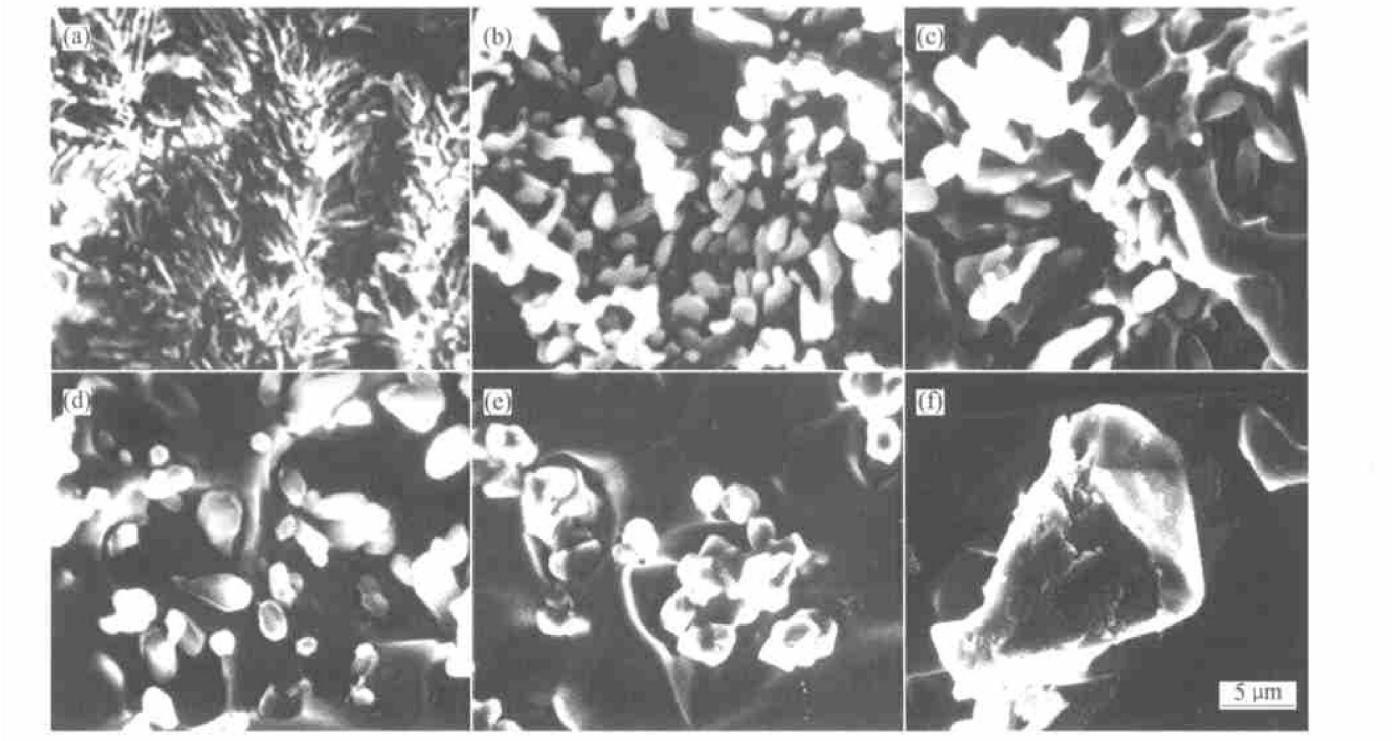

Fig. 3 Morphology evolution of eutectic silicon of alloy during solution treatment at 525 ℃ for various times
(a) —0 h; (b) —2 h; (c) —6 h; (d) —10 h; (e) —14 h; (f) —24 h

Besides composition homogeneity by solution treatment, the evolution of eutectic silicon morphology is another reason to influence the strengthening of the alloy by solution treatment. Due to the thermal effect, the unstable eutectic silicon modified by sodium-salts dissolves and fragmentizes. In the meantime, a tendency of spheroidization happens due to the driving force of irregular curvature. The mechanics field can be improved between silicon particles and matrix for the evolution of silicon morphology[5], which enhances the mechanical properties of alloy.

3.3 Quantitative metallographic analysis of eutectic silicon

In order to quantitatively describe the evolution of silicon morphology during solution treatment, the quantitative metallgraphic analysis of silicon particles as a function of solution time at 525 ℃ is shown in Fig. 4. During solution treatment, the mean diameter of silicon particles increases, but the silicon particle counts per unit area, area ratio and length/width ratio tend to decrease with the increase of solution time. According to the features of curves in Fig. 4, it can be classified into three stages: initial, intermediate and final stages.

In the early stage of solution treatment(1 - 6 h), the increase of mean diameter of silicon particles, and the decrease of silicon particle counts per unit area as well as area ratio are remarkable, but length/width ratio reduces slowly. Because of large diffusion coefficient at 525 ℃ and large concentration gradient of sil-

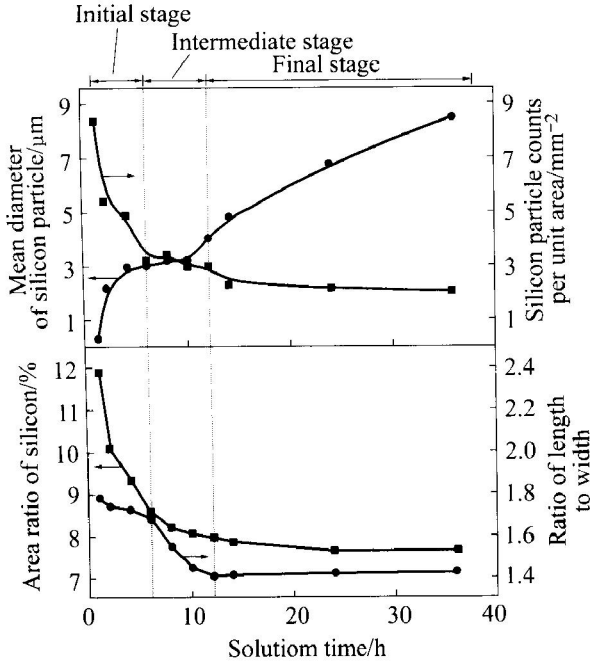

Fig. 4 Quantitative metallographic parameters of silicon particles as function of solution time at 525 ℃

icon between eutectic silicon and matrix. silicon atoms diffuse very quickly in aluminum matrix and a part of silicon atoms have diffused and dissolved into matrix in a large solubility at 525 ℃. Necking, fragment and growth of silicon particles firstly occur due to homogeneous energy caused by atoms misalignment, crystal defect and rounding tendency of the tips of particles, that results in the increase of mechanical properties, especially elongation, and the fracture surface transforms to equiaxed dimple.

In the intermediate stage (6 – 12 h), the decrease of silicon particle counts per unit area and area ratio of silicon become gentle, but the ratio of length to width of silicon particles reduces quickly. This behavior indicates that the fragmentation of silicon particles has finished and spheroidization is going on. The diffusion and dissolution tend to balance in this stage, for the content of silicon is saturated in matrix. At the combined action of heat and curvature, silicon particles tend to regularize and spheroidize in order to reduce the surface energy. Maximum hardness(strength) have attained after solution treated for 10 – 12 h due to perfect morphology, suitable size and uniform distribution of eutectic silicon, as well as the age-hardening of Al_2Cu and Mg_2Si which have completely dissolved in matrix during solution treatment.

In the final stage (12 – 36 h), except for mean diameter of silicon particles, the change of other parameters is not obvious. The decrease of hardness and strength is related to the deterioration of silicon morphology (Fig. 3(f)). The growth of silicon phase is a coarsening process controlled by diffusion. The silicon atoms attach to particle surface directly and preferentially. As a result, the facets and lap occur in silicon particles[5]. The coarsening process of silicon follows LSW(Liftshitz-Slyozov-Wagner) model[11], having a linear relationship between the solution time and the cubic of mean diameter, as shown in Fig. 5.

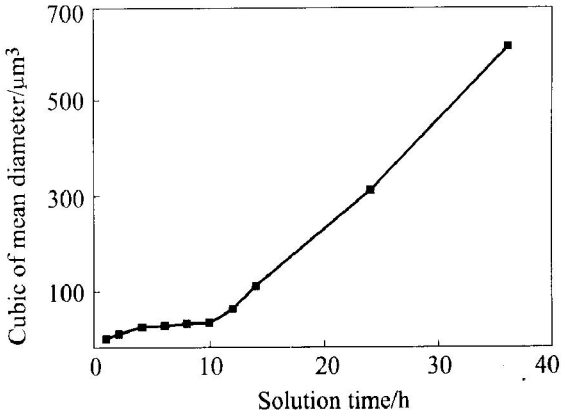

Fig. 5 Coarsening process of silicon particles with solution time

3.4 Precipitation of excess silicon in aging treatment

Fig. 6(a) shows the TEM micrograph of the alloy solution treated at 525 ℃ for 12 h and then aged at 175 ℃. Few precipitates can be observed in as-quenched sample, while after aged for 6h besides those fine GP zones (or β″ and θ″ phases) some precipitates (shown by arrow) with size of 100 – 200 nm are observed, which are proved to be Si phase by diffraction pattern (Fig. 6(b)). Fig. 7 shows the age-hardening curve of the alloy at 175 ℃, where obvious age-hardening phenomenon can be found and the peak ageing is obtained at 6 h and then the hardness drops because of over-ageing.

As shown by Zhang's measurement using electron microprobe analysis[7], as much as 1.3% Si (mass fraction) can be retained in the solid solution of the matrix after Al-7% Si-0.4% Mg(mass fraction) casting alloy is solution treated at 540 ℃ and water quenched. In present case, the silicon content in α (Al) solid solution is approximately 1.3%(mass fraction), but silicon solubility at room temperature is only 0.05%[5]. With 0.43% Mg in the matrix, excessive content of Si is needed to balance the Mg content to form Mg_2Si compound during aging in which mole ratio of Mg to Si is 1.73. Besides β″ and θ″ phases, it has been observed that some silicon precipitates exist in microstructure of peak aged alloy, as shown in Fig. 5(a). For this condition, the excess silicon may be present as clusters of silicon atoms. These clusters may be too small to be observed using conventional TEM techniques, but they could work as barriers for dislocation movement, and thus enhance the strength of alloy[7, 10]. A higher level of excess silicon may not only cause formation of finer β″ precipitates because of its effect on nucleation, but al-

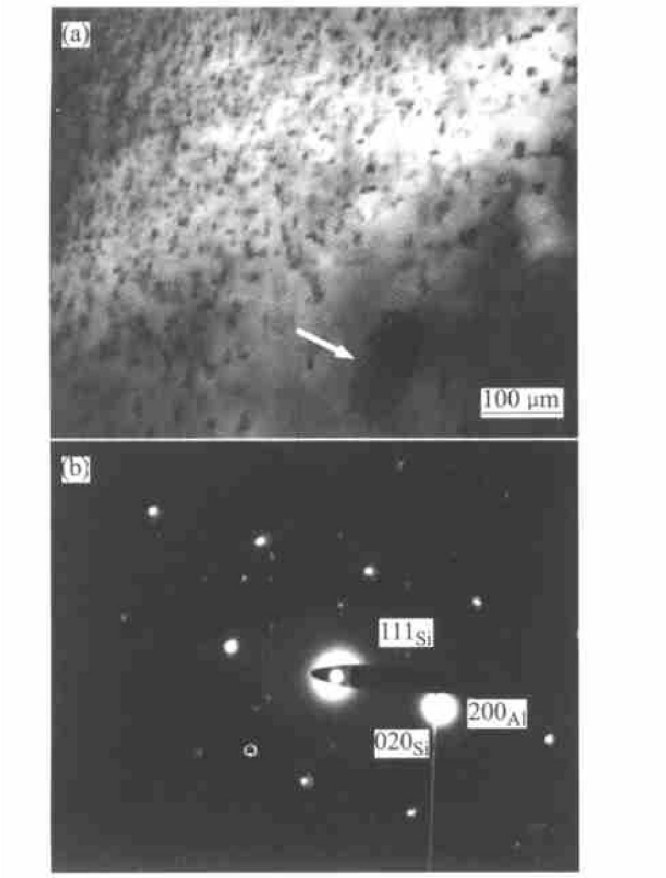

Fig. 6 TEM micrograph(a) and corresponding SADP(b) of alloy aged at 175 ℃

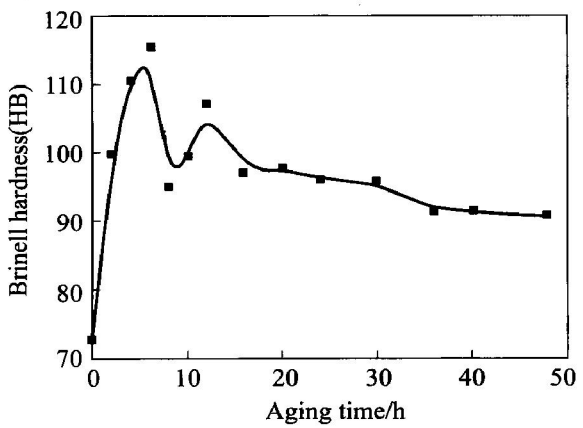

Fig. 7 Age-hardening curves of alloy aged at 175 ℃

so result in a higher total volume fraction of β″ precipitates[10]. In present study, it is found that the precipitation rate and growth of silicon are larger than those of β″ and θ″ phase. This behavior may result in acceleration of over-aging for the alloy, indicating that silicon exerts an essential and comprehensive effect on the precipitation hardening process of the alloy[11-15].

REFERENCES

[1] HUAN Liang-yu. The development of Al-Si alloy and new viewpoint[J]. Special Casting & Nonferrous Alloys, 1995(4): 30-32. (in Chinese)

[2] XUN Yan-cai, LIU Bo-cao. Present status and future development of cast aluminum alloy[J]. Special Casting & Nonferrous Alloys, 1998(4): 1-5. (in Chinese)

[3] LI Rong-de, YU Hai-peng, GU Ke-cheng, et al. Quantitative metallographic analysis for morphology change of eutectic silicon in Al-Si-Cu-Mg hypoeutectic casting alloy during the solution heat treatment[J]. Foundry, 1997(6): 1-5. (in Chinese)

[4] Meyers W. Solution heat treatment effects in A357 alloys [J]. AFS Trans, 1985, 112: 741-750.

[5] SUN Yu, CHEN Jin, SUN Guo-xiong. Evolution of Si phase in Al-Si alloy and its effect on mechanical properties [J]. Special Casting & Nonferrous Alloys, 2001(6): 1-4. (in Chinese)

[6] Pedersen L. The effect of solution heat treatment and quenching rates on mechanical properties and microstructure in AlSiMg foundry alloys[J]. Metall Mater Trans A, 2001, 32A(3): 525-532.

[7] Zhang D L. Precipitation of excess silicon during heat treatment of cast Al-7wt% Si-0.4wt% Mg alloy[J]. Materials Science Forum, 1996, 217-222: 771-776.

[8] Kashyap K T. Casting and heat treatment variables of Al-7Si-Mg alloy[J]. Mater Sci Technol, 1993, 9(3): 189-203.

[9] Kaczorowski M. The role of excess silicon in the process of precipitation hardening of AlSiMg alloy (Ⅰ)[J]. Aluminum, 1985, 10: 756-762.

[10] Zhang D L, Zheng L. The quench sensitivity of cast Al-7wt% Si-0.4wt% Mg alloy[J]. Metall Mater Trans A, 1996, 27A: 3983-3391.

[11] Kang H G. Age-hardening characteristics of Al-Si-Cu-base cast alloy[J]. AFS, 1999, 27: 507-516.

[12] LIAO Heng-chang, SUN Guo-xiong. Interaction between Sr and B in Al-Si casting alloys[J]. Trans Nonferrous Met Soc China, 2003, 13(2): 353-361. (in Chinese)

[13] GUI Man-chang, JIA Jun, Li Qing-chun. Effect of Cd addition on the microstructure and properties of Al-Si_7Cu_2Mg alloy[J]. The Chinese Journal of Nonferrous Alloys, 1994, 4(2): 69-75. (in Chinese)

[14] Schueller R D, Wawner F E. Nucleation mechanism of the cubic σ phase in sequeeze-cast aluminium matrix composites[J]. Journal of Materials Science, 1994, 29: 424-236.

[15] LI Rong-de, YANG Xiu-ying, LI Run-xia. Computer simulation on primary α-phase of Al-7% Si solidification structure[J]. Trans Nonferrous Met Soc China, 2003, 13(5): 1128-1132.

(Edited by YANG Bing)