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ScienceDirect
Trans. Nonferrous Met. Soc. China 23(2013) 2518−2523
Transactions of Nonferrous Metals Society of China
www.tnmsc.cn

Effect of Mg and semi solid processing on microstructure and impression creep properties of A356 alloy

B. AMIR ESGANDARI[1], B. NAMI[2], M. SHAHMIRI[1], A. ABEDI[2]

1. School of Metallurgy and Materials Engineering, Iran University of Science and Technology, Tehran, Iran;
2. Department of Mechanical Engineering, Shahid Rajaee Teacher Training University, Tehran, Iran

Received 1 November 2012; accepted 2 May 2013

Abstract: The effects of Mg and semi solid processing on the creep properties of A356 Al alloy were investigated. The results show that the dislocation climb controlled creep is the dominant creep mechanism and it is not affected by the semi solid processing and further addition of Mg. Mg improves the alloy creep properties probably by forming large Chinese script Mg_2Si compounds at the interdendritic regions. The semi solid processed specimens exhibit better creep properties in comparison with the as cast ones. It is attributed to the reduction in the stacking fault energy resulting from the significant dissolution of Mg in the α(Al) phase.

Key words: A356 alloy; microstructure; semi solid processing; impression creep; dislocation climb controlled creep

1 Introduction

Among the Al based casting alloys, Al−Si−Mg alloys are the most widely used alloys in aerospace and automotive industries because of the good casting properties accompanied by the superior mechanical properties [1]. Mg as an alloying element can enhance the mechanical properties of Al−Si−Mg casting alloys due to precipitating Mg_2Si phase during T6 treatment [2]. On the other hand, although higher amount of Mg content raises strength of the alloys, the formation of large Chinese script Mg_2Si compound during the solidification deteriorates their ductility and fracture toughness [3]. Results of previous studies have shown that semi solid processing modifies the microstructure and enhances the mechanical properties [4,5]. The shrinkage porosity in the semi solid processed products is very low in comparison with that produced by conventional casting method, resulting in the enhanced mechanical properties [6].

Many studies have been carried out on the evaluating microstructure and room temperature mechanical properties of A356 alloy produced by semi solid processing [7−9]. However, a little is known on the effect of alloying elements and semi solid processing on the creep behavior of the alloy. Therefore, the aim of this work is to investigate the effect of Mg and semi solid processing on the microstructure and impression creep properties of A356 alloy.

2 Experimental

Two alloys based on Al−Si alloys with different amounts of Mg were used in this work. The chemical compositions of the alloys are listed in Table 1. The alloys were melted in a graphite crucible at 750 ºC and poured in a permanent steel mould heated to about 200 °C. Cylindrical slices with 20 mm diameter and 10 mm height were machined from the cast alloys. The semi-solid microstructure was obtained using recrystallization and partial remelting (RAP) process. Then the specimens were compressed about 30% with a hydraulic press machine at ambient temperature and were partially remelted in a salt bath at (590±2) °C for 1 h. The partially remelted specimens were subsequently quenched in water bath at room temperature.

The slices with 6 mm thickness were prepared from the partially remelted and quenched specimens. Creep test was performed on the specimens using the impression creep technique at the temperature range of 150 to 212 °C under the shear modulus-normalized stress (σ/G) between 0.0225 and 0.035 for 75 min. The samples for microstructure investigation were prepared by usual

Corresponding author: B. NAMI; Tel/Fax: +982122970052; E-mail: bnami@srttu.edu
DOI: 10.1016/S1003-6326(13)62763-6

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

Alloy	Si	Mg	Fe	Cu
A356	7.31	0.32	0.11	0.03
A356+1%Mg	7.22	1.35	0.1	0.02
Alloy	Mn	Zn	Ti	Al
A356	0.03	0.027	0.045	Bal.
A356+1%Mg	0.03	0.024	0.041	Bal.

standard grinding and polishing methods. Microstructure characterization was performed by an optical microscope and a LEO-1550 scanning electron microscope (SEM) equipped with an energy dispersive spectroscopy (EDAX). Phase identification was examined by X-ray diffraction (XRD).

3 Results and discussion

Figure 1 shows the optical microstructure of as cast A356 and A356+1%Mg alloys. According to Fig. 1 and previous work [5], the microstructure of A356 alloy is composed of α(Al) phase and an eutectic phase in the interdendritic regions including α(Al) and Si phases. In the A356 alloy, α(Al) phase has dendritic morphology. Comparing the microstructures of the alloys, it is indicated that adding Mg to A356 alloy changes the dendritic microstructure to an equiaxed microstructure. The eutectic Si has needle shape morphology in both the alloys and adding Mg has no obvious effect on the morphology.

Figure 2 shows SEM images of the as cast A356 and A356+1%Mg alloys. As seen, the amount of Mg_2Si phase with Chinese script morphology in A356+1%Mg alloy is greater than that in A356 alloy.

Optical microstructures of both the semi solid processed A356 and A356+1%Mg alloys are illustrated in Fig. 3. It is obvious that the morphology of α(Al) phase is completely changed to a globular and spherical shape. It is observed that the average size of globular α(Al) phase in A356+1%Mg alloy is a little smaller than that in A356 alloy. Comparing Figs. 3(b) and (d), it is revealed that the eutectic silicon is modified in both alloys and its size is smaller in A356+1%Mg alloy than that in A356 alloy.

Figure 4 shows the impression creep curves of A356+1%Mg alloy under different applied stresses. The creep behaviors are expressed as the impression depth versus time. Almost all the curves exhibit a primary creep regime followed by a steady-state creep. In the primary creep stage, work hardening rate is more than recovery rate, which results in a decrease in the creep rate with time. Work hardening and recovery rates reach equilibrium during the secondary creep stage, resulting in the constant creep rate. Although tertiary creep resulting from the specimens necking and forming creep voids are observed in the conventional tensile creep curves, they are not observed in the impression creep curve. However, the presence of tertiary creep during impression creep is reported in Ref. [10] due to

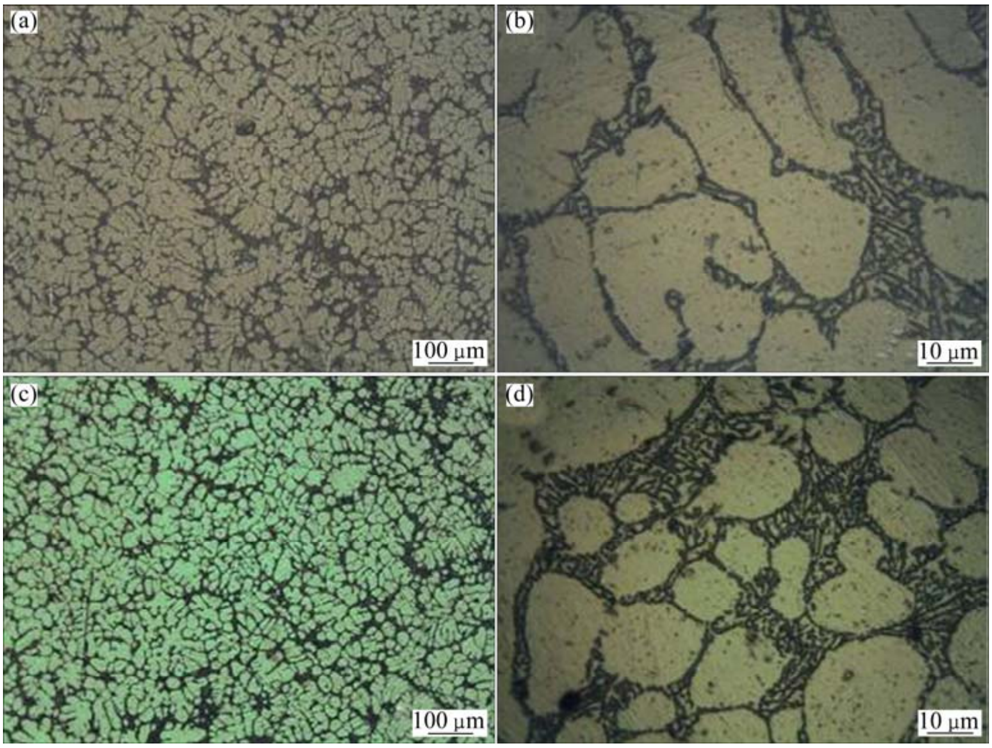

Fig. 1 Optical microstructures of as cast specimens: (a, b) A356 alloy; (c, d) A356+1% Mg alloy

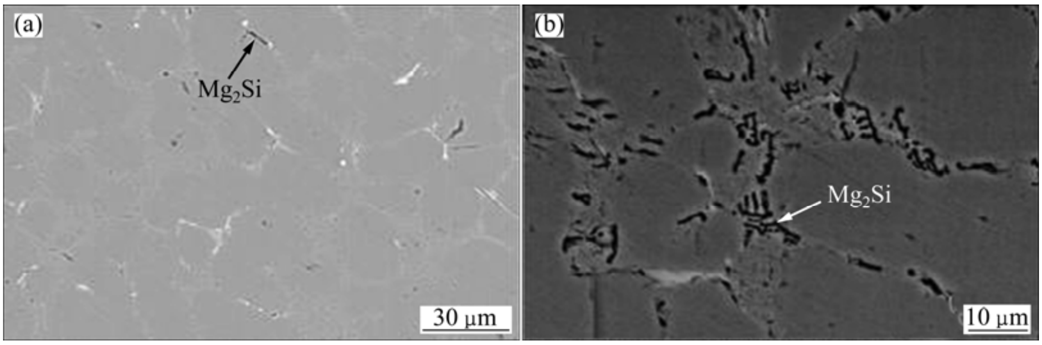

Fig. 2 SEM images of as cast specimens: (a) A356 alloy; (b) A356+1%Mg alloy

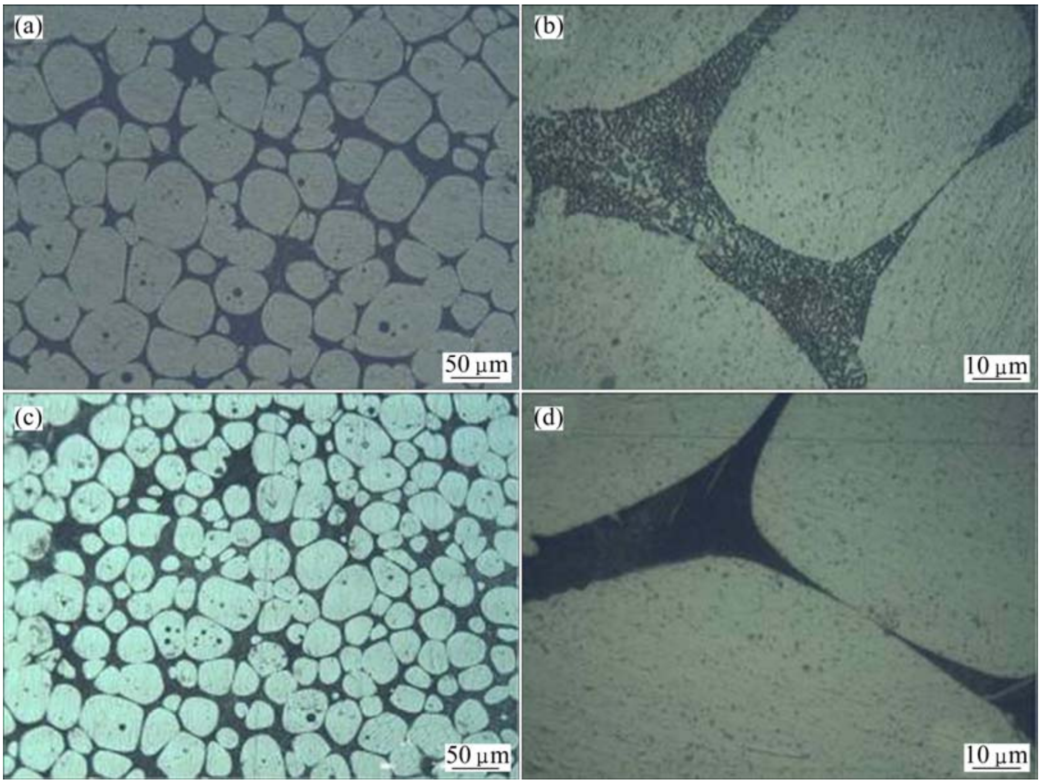

Fig. 3 Optical microstructures of semi solid processed specimens: (a, b) A356 alloy; (c, d) A356+1%Mg alloy

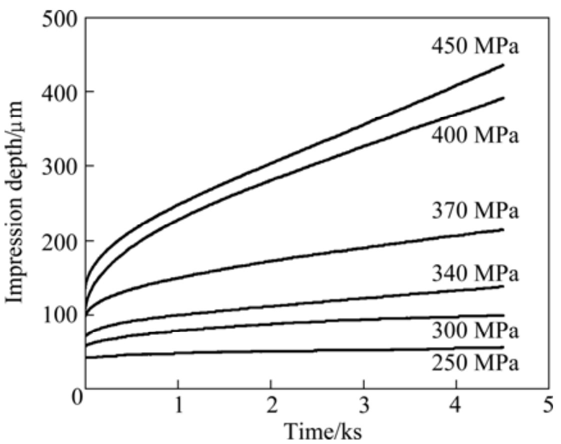

Fig. 4 Variation of impression depth with creep time at different applied stresses for semi solid processed A356+1%Mg alloy

decreasing creep strength of specimens with time. Therefore, the absence of tertiary creep in Fig. 4 indicates that the alloys preserve their creep resistance during the creep test.

It has been reported that the variation of steady state impression creep velocity (v_{imp}) versus impression stress (σ) and testing temperature T can be expressed using the following power law equation [11]:

$$v_{imp} = A\sigma^n \exp\left(-\frac{Q}{RT}\right) \tag{1}$$

where Q is the creep activation energy, R is the gas constant and A is the material-dependent constant.

Figure 5 shows the variation of v_{imp} versus σ in logarithmic axes for both alloys produced by both conventional and semisolid casting processes. It is observed that the creep resistance of the conventionally cast A356+1%Mg alloy is slightly higher than that of A356 alloy. Similar to the conventionally cast specimens, the semi solid processed A356+1%Mg alloy has slightly better creep properties than the A356 alloy. Improving the creep resistance of the semi solid processed specimens is also obvious in Fig. 5.

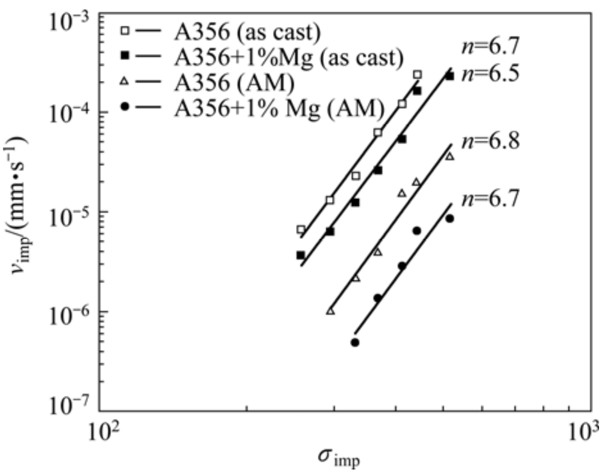

Fig. 5 Variation of impression velocity with applied stress for A356 and A356+1%Mg produced by conventional and semi solid casting processes

According to Ref. [12], determining the creep mechanism can help to explain the improvement reasons for the alloy creep resistance resulting from adding Mg and semi solid processing. As Eq. (1) reveals, the value of stress exponent n, demonstrating the creep mechanism is obtained from the slope of the variation of $\ln(v_{imp})$ versus $\ln(\sigma)$ at constant temperature. Comparing the obtained stress exponent values shown in Fig. 5 with those reported in Ref. [10,11] indicates that the climb controlled dislocation creep is the dominant mechanism during creep of the as cast and semi solid processed alloys. The similar obtained values of n for all cases imply that the creep mechanism is not affected by adding Mg and processing technology.

Creep improvement during dislocation climb controlled creep may be taken place by three factors: 1) formation of small precipitates with high temperature stability inside the grains; 2) large intermetallic components between the grains and interdendritic regions; 3) decreasing stacking fault energy of matrix phase [13−16]. The small particles improve the creep resistance by impeding easy movement of dislocation inside the grains. The large intermetallic compounds increase the creep properties by reducing annihilation of dislocation in the grain boundaries and by decreasing grain boundary sliding. The distance between partial dislocations increases by decreasing stacking fault energy, resulting in the difficulty of dislocation climbing. In such cases, greater stress is needed for dislocations to combine with together and climb over the obstacles. Although Mg addition has no influence on the morphology of Si, large Chinese script Mg_2Si compounds are formed in the interdendritic regions, as shown in Fig. 2. On the other hand, EDS analysis of α(Al) in Table 2 shows similar concentration for this phase in both alloys. It can be seen that Mg has no considerable solubility in the matrix. Therefore, the formation of large Mg_2Si compounds might be the sole factor accounting for the better creep resistance of the as cast A356+1%Mg alloy in comparison with A356 alloy.

Table 2 EDS analysis of α(Al) in conventionally cast A356 and A356+1%Mg alloys

Alloy	Composition/%		
	Al	Si	Mg
A356	99.2	0.8	−
A356+1%Mg	99.1	0.9	−

Figure 5 indicates that the semi solid processed specimens have better creep properties comparing to the conventional cast specimens. According to the XRD pattern of the semisolid processed A356+1%Mg alloy in Fig. 6, Mg_2Si does not present in the microstructure of alloy. Even though, the absence of Mg_2Si phase is predicted to decline the creep properties of alloy. Therefore, the improvement in creep properties of the semi solid processed specimens contrary to the as cast specimens can not be attributed to the forming of large intermetallic compounds. Table 3 shows the results of EDS analysis of α(Al) phase in the semi solid processed specimens of both alloys. It is observed that considerable

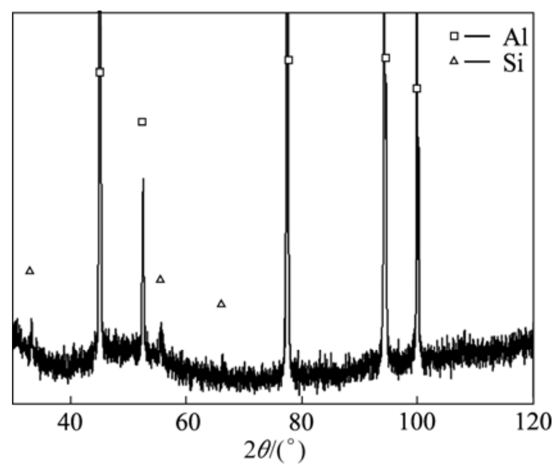

Fig. 6 XRD pattern of A356+1%Mg alloy produced by semi solid casting process

Table 3 EDS analysis of α(Al) in semi solid processed A356 and A356+1% Mg alloys

Alloy	Composition/%		
	Al	Si	Mg
A356	98.88	0.95	0.17
A356+1%Mg	98.54	1.07	0.39

amount of Mg is dissolved in α(Al) phase. Therefore, it seems that Mg reduces stacking fault energy of α(Al) phase, resulting in the increased creep properties. Because of more dissolved Mg in α(Al) phase of A356+1%Mg comparing to that in α(Al) phase of A356, more reduction in stacking fault energy occurs in A356+1%Mg alloy. Therefore, A356+1%Mg alloy exhibits better creep resistance in comparison with A356 alloy.

4 Conclusions

1) Adding Mg to conventionally cast A356 alloy causes to forming large Mg_2Si intermetallic component with Chinese script morphology in the interdendritic regions. Mg has no considerable effect on the morphology of eutectic Si in the conventionally cast specimens.

2) The morphology of α(Al) phase is changed to a globular microstructure in the semi solid processed specimens and eutectic Si is modified.

3) Calculation of stress power (n) indicates that the climb controlled dislocation creep is the dominant creep mechanism. Adding Mg and semi solid processing have no influence on the creep mechanism.

4) A356+1%Mg alloy exhibits better creep resistance than A356 alloy. It is because of the presence of large Chinese script Mg_2Si intermetallic compound in the microstructure of the alloy which reduces annihilation of dislocation and sliding of grain boundary.

5) Semi solid processed specimens show better creep resistance than the as cast ones. It is probably due to dissolution of Mg in α(Al) phase and subsequent reducing in stacking fault energy.

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Mg 和半固态加工对 A356 合金显微组织和抗蠕变性能的影响

B. AMIR ESGANDARI[1], B. NAMI[2], M. SHAHMIRI[1], A. ABEDI[2]

1. School of Metallurgy and Materials Engineering, Iran University of Science and Technology, Tehran, Iran;
2. Department of Mechanical Engineering, Shahid Rajaee Teacher Training University, Tehran, Iran

摘　要：研究 Mg 和半固态加工对 A356 合金蠕变性能的影响。结果表明：位错攀移控制的蠕变是占主导地位的蠕变机制，它不会受到半固态加工以及进一步添加 Mg 的影响。Mg 提高合金的抗蠕变性能主要是由于在枝晶间区域形成大量汉字型的 Mg_2Si。半固态加工的样品表现出比铸造样品更好的抗蠕变性能，这是由于 Mg 在 α(Al)相中显著溶解所致堆垛层错能的减少而导致的。

关键词：A356 合金；显微组织；半固态加工；压入蠕变；位错攀移控制蠕变

(Edited by Chao WANG)