

Article ID: 1003 - 6326(1999)03 - 0500 - 05

## High strain rate superplasticity of SiC whisker reinforced pure aluminum composites<sup>①</sup>

Xu Xiaojing(许晓静), Zhao Changzheng(赵昌正), Zhang Di(张荻),  
Shi Zhongliang(施忠良), Wu Renjie(吴人洁)  
State Key Laboratory of Metal Matrix Composites,  
Shanghai JiaoTong University, Shanghai 200030, P. R. China

**Abstract:** A  $\beta$  SiC whisker reinforced pure aluminum composites expected to exhibit high strain rate superplasticity has been successfully fabricated by a new processing route consisting of pressure infiltration, extrusion with a low extrusion ratio and rolling. The composites exhibit a total elongation of 220% ~ 380% in the initial strain rates within  $1.0 \times 10^{-2} \sim 1.0 \times 10^{-1} \text{ s}^{-1}$  and at 893 ~ 903 K. According to differential thermal analysis(DTA) and microstructure observation, it is concluded that an appropriately small amount of liquid phase is necessary to cause a good high strain rate superplasticity in aluminum matrix composites in addition to fine and uniform microstructure.

**Key words:** high strain rate; superplasticity; aluminum matrix composites; pressure infiltration

**Document code:** A

### 1 INTRODUCTION

Some fine ceramic whisker or particle reinforced aluminum matrix composites can exhibit a total elongation of 200% ~ 600% at a high strain rate ( $> 10^{-2} \text{ s}^{-1}$ )<sup>[1-8]</sup>. The high strain rate superplasticity (HSRS) is very attractive for commercial applications because one of the major drawbacks in the current superplastic forming technology is slow forming rates of typically  $10^{-5} \sim 10^{-3} \text{ s}^{-1}$ .

So far, the HSRS composites, which can exhibit a maximum total elongation ( $> 300\%$ ), have been fabricated primarily by powder metallurgy and hot extrusion (or hot extrusion + hot rolling) with a very large extrusion rate of typically 50:1 ~ 100:1<sup>[1-8]</sup>. But this processing route is cost-expensive and therefore is not desirable for industrial fabrication.

It has been widely accepted that the optimum temperature for HSRS in aluminum matrix composites is close to or just slightly above the solidus temperature of the aluminum matrix, so

it is possible that during the superplastic deformation there is some liquid phase presented in the composites. Currently the effect of the liquid phase on the HSRS behaviors is the subject of great interest<sup>[9,10]</sup>. In an aluminum matrix composites, solute atoms can have a great influence on its melting behavior and further on its superplastic behavior, it is, therefore, necessary to investigate the simple superplastic characteristics of pure aluminum matrix composites in order to examine the role of liquid phase and the deformation mechanisms for the HSRS. Recently, Imai *et al* reported the observation of HSRS in some pure aluminum matrix composites, their PM SiC<sub>p</sub>/1100 (99.84% Al), PM SiC<sub>p</sub>/IN90 (99.99% Al) and SQ  $\beta$  Si<sub>3</sub>N<sub>4</sub>/99.99% Al composites, after extruded (extrusion ratio 100:1) and rolled, can exhibit HSRS, but the maximum total elongation was only 200%<sup>[7,8]</sup> (note: PM, SQ represents powder metallurgy method and squeeze casting method, respectively).

The purposes of this study are to report the

① Project 59631080 supported by the National Natural Science Foundation of China

Received Jun.16, 1998; accepted Oct.24, 1998

observation of HSRS in an SiC<sub>w</sub>/ pure aluminum composites made by pressure infiltration and rolling at 623 K after extrusion with a low extrusion ratio of 10:1, and then to examine the role of liquid phase and the superplastic deformation mechanisms.

## 2 MATERIALS AND EXPERIMENTAL PROCEDURE

The reinforcement materials used in this investigation were  $\beta$  SiC whiskers, the diameter is about 0.1 ~ 1.0  $\mu\text{m}$  and the length/diameter ratio is about 2 ~ 10. The whisker preform was fabricated by a water absorbed method. The matrix (pure aluminum) had a chemical composition of Al-0.40Si-0.35Fe-0.05Cu (%). After the preform and the pure aluminum were respectively heated up to 923 K and 1023 K, the molten pure aluminum was infiltrated into the preform under the pressure of about 9 MPa, then condensed at the same pressure. The volume fraction of the SiC whisker in the composites was  $\varphi_f = 0.2$  after infiltration.

Thermomechanical processing for HSRS included hot extrusion with a low extrusion ratio of 10:1 at 723 K and further hot rolling at 623 K. Rolling reduction per pass was 10% and the reheating time between rolling passes was about 5 min. The final thickness of the hot-rolled composites sheet was about 0.4 mm (The total rolling reduction was about 86%).

The superplastic tensile specimens with 4 mm  $\times$  5 mm were directly machined from the rolled composites sheet in such a way that the test direction was parallel to the rolling direction. Prior to testing, the specimen was held at the specified test temperature for about 20 min to establish thermal equilibrium. Specimens were pulled in air at constant crosshead speed.

## 3 RESULTS AND DISCUSSION

### 3.1 Distribution of SiC whiskers

The uniform distribution of fine ceramic reinforcements is vital to cause HSRS in aluminum matrix composites. Fig. 1 shows the optical structure of the rolled composites. The distribu-

tion of SiC whiskers is very homogeneous and approximately aligned in the direction of rolling direction.

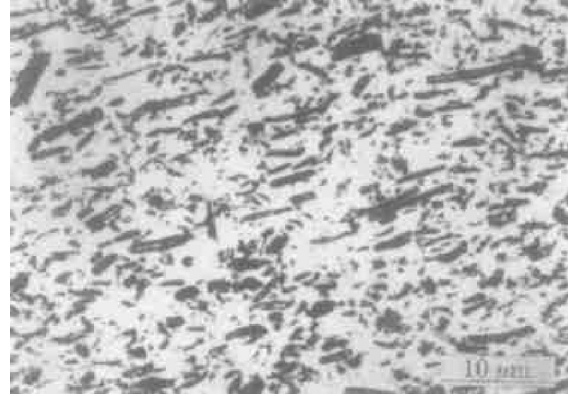


Fig.1 Optical structure of rolled SiC<sub>w</sub>/ pure Al composites

### 3.2 DTA analysis

Differential thermal analysis (DTA) with a constant heating rate of 5 K/min was performed to determine the exact melting behavior of the composites as shown in Fig. 2. The macroscopic melting temperature ( $T_3$ ) of the composites is 928.7 K, the temperature difference decreases at two different temperature points, first at  $T_1$  (888.1 K), then at  $T_2$  (about 913.0 K).

Fig.2 DTA analysis for rolled SiC<sub>w</sub>/ pure Al composites

$T_1$  — Incipient melting temperature, 888.1 K;

$T_2$  — Whole grain boundary and interface melting temperature, 913.0 K;

$T_3$  — Macroscopic melting temperature, 928.7 K

Macroscopic melting temperature ( $T_3$ ) is

the nominal temperature at which the matrix (i.e. grain inside) begins to melt.

$T_1$  is called as incipient melting temperature. It has been revealed that the incipient melting is a result of the segregation of solute atoms (such as Cu, Mg and Si) in some local areas at the reinforcement-matrix interfaces and grain boundaries because such segregation decreases the local melting point at interfaces and grain boundaries<sup>[11,12]</sup>. It can be seen from Fig. 2 that in the temperature region from  $T_1$  to  $T_2$ , the decrease in the temperature-difference is small and slow, indicating that there is only a small amount of liquid phase presented in the composites.

Up to now, the report on what physical phenomenon is concerned with the  $T_2$  has not been found yet, but from the observation that beginning just from  $T_2$  the temperature-difference decreases quickly and continuously as shown in Fig. 2, indicating that the liquid phase increases quickly and continuously, it can be inferred that  $T_2$  is the temperature at which the interfaces and the grain boundaries wholly begin to melt, and that at a slightly higher temperature than  $T_2$ , the whole interfaces and grain boundaries would be melted completely.

### 3.3 Superplastic deformation characteristics

According to the DTA results above, four temperatures (883 K, 893 K, 903 K and 913 K) were selected as the superplastic testing temperature in order to examine the role of liquid phase in the superplasticity of the composites.

The relationship between flow stress ( $\sigma$ ) and true strain rate ( $\dot{\epsilon}$ ) in a superplastic material can be expressed by the following empirical equation:

$$\sigma = k\dot{\epsilon}^m$$

where  $k$  is a constant and  $m$  is the strain rate sensitivity value defined by the slope of the curve of  $\ln \sigma$  vs  $\ln \dot{\epsilon}$ . It is necessary to find the optimum superplastic conditions in which the  $m$ -value is greater than 0.30 because a high  $m$ -value would suppress neck formation and leads to high elongation.

Fig.3 shows the relationships between the

flow stress and the true strain rate of the composites at a true strain (0.2). At 893 K and 903 K, the  $m$ -values are high, about 0.44 and 0.52, respectively. But at 883 K and 913 K, the  $m$ -values decrease rapidly to be 0.21 and 0.16, respectively.

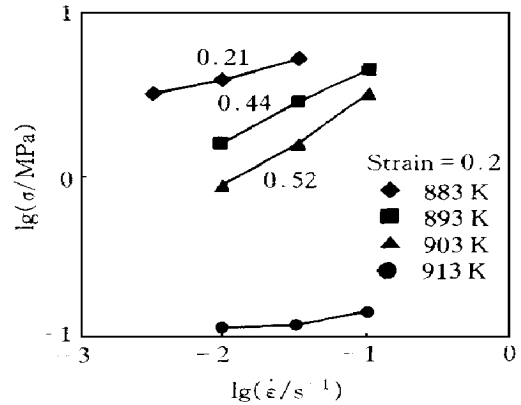


Fig.3 Relationships between flow stress and strain rate of SiC<sub>w</sub>/pure Al composites

Fig.4 shows the relationships between the total elongation and the initial strain rate ( $\epsilon_0$ ) of the composites. In the wide initial strain rate region of  $1.0 \times 10^{-2} \sim 1.0 \times 10^{-1} \text{ s}^{-1}$  and at the temperatures of 893 ~ 903 K, the composites exhibit a total elongation of 220% ~ 380%. A maximum total elongation of 380% is obtained at the initial strain rate of  $3.3 \times 10^{-2} \text{ s}^{-1}$  and at 903 K, Fig.5 shows the comparison of the untested and deformed composites. However, at 883 K and 913 K, the total elongations are low, less than 160% as shown in Fig.4.

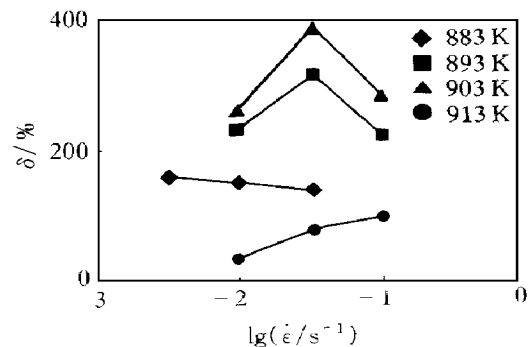
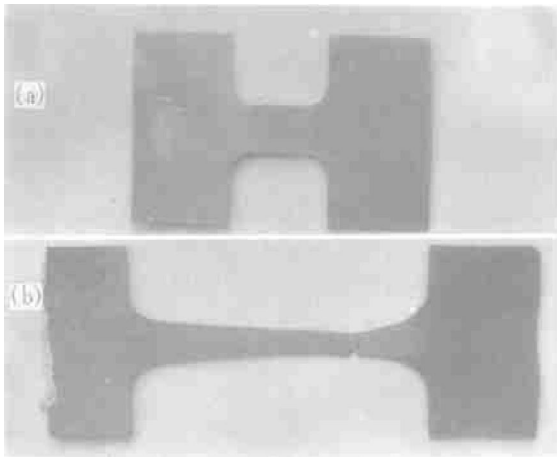


Fig.4 Relationships between total elongation and strain rate of SiC<sub>w</sub>/pure Al composites



**Fig. 5** A comparison of untested and superplastically deformed specimens  
(a) — Before test ; (b) —  $\dot{\epsilon}_0 = 0.033 \text{ s}^{-1}$ ,  
 $T = 903 \text{ K}$ , elongation = 380 %

To make clear the role of liquid phase in the HSRS, the relationship between the total elongation and the superplastic temperatures at an initial strain rate of  $3.3 \times 10^{-2} \text{ s}^{-1}$  is shown in Fig. 6. It is noted that the SiC<sub>w</sub>/pure Al composites can exhibit a good HSRS only in the narrow temperature within 893 ~ 903 K, which is 5 ~ 15 K higher than the  $T_1$  (incipient melting temperature) but is 20 ~ 10 K lower than the  $T_2$  (whole interfaces and grain boundaries melting

**Fig. 6** Relationships between elongation and temperature of SiC<sub>w</sub>/pure Al composites

$T_1$  — Incipient melting temperature ;

$T_2$  — Whole interfaces and grain boundaries melting temperature ;

$T_3$  — Macroscopic melting temperature

temperature), indicating that liquid phase plays a vital role to high strain rate superplasticity of aluminum matrix composites, an appropriate amount of liquid phase can enhance the HSRS to a great extent and a less or slightly more would decrease the total elongation quickly.

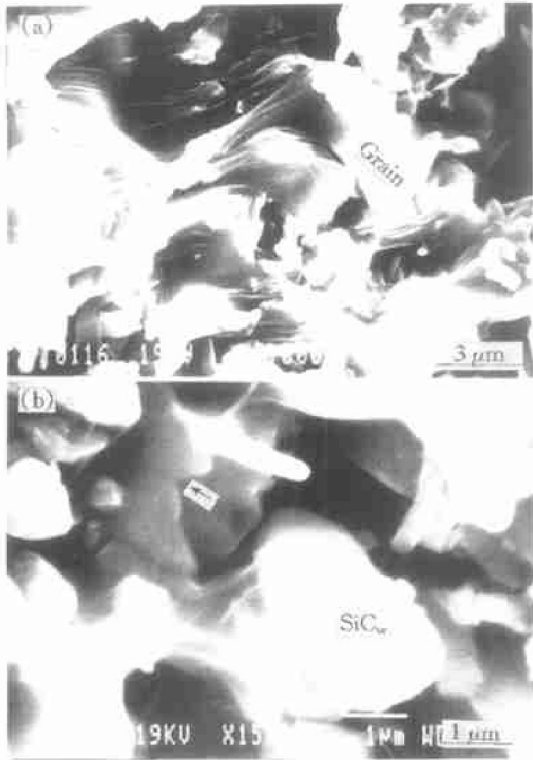
During superplastic deformation of the composites containing hard reinforcements, stress concentrations are often caused at the reinforcement/matrix interfaces, an appropriate amount of liquid phase, due to helping to relieve the stress concentration without excessive damage to the bonding force between grain boundaries (and interfaces), and consequently can enhance the superplasticity. But when there is too much liquid phase presented at grain boundaries and interfaces, grain boundaries and interfaces can no longer support an applied tensile force, as a result, the total elongation would decrease quickly.

Fig. 7 shows the fracture surface of the composites pulled at the optimum superplastic condition (i.e.  $3.3 \times 10^{-2} \text{ s}^{-1}$ , 903 K, 380%). Fig. 7(a) shows that a lot of “ripples” appear at grain boundaries, which is the trace left behind by grain boundary sliding<sup>[13]</sup>, this indicates that extensive fine grain boundary sliding has taken place during the superplastic deformation of the composites (grain size is about 1 ~ 4 μm). Fig. 7(b) shows whisker pull-out and good grain boundary bonding (arrowed in Fig. 7(b)). The whisker pull-out indicates that interfacial sliding has also taken place during the superplastic deformation of the composites. The good grain boundary bonding indicates that at 903 K the grain boundaries have not been melted completely and have some bonding force.

In conclusion, the primary mechanism for HSRS in the composites is fine grain boundary sliding and interfacial sliding accommodated primarily by an appropriately small amount of liquid phase.

#### 4 CONCLUSIONS

(1) The as-processed composites exhibit a strain rate sensitivity ( $m$  value) of about 0.44 ~ 0.52 and a total elongation of 220% ~ 380%



**Fig.7** SEM images of fracture surface of SiC<sub>w</sub>/pure Al composites after superplastic deformation at  $3.3 \times 10^{-2} \text{ s}^{-1}$  and 903 K

- (a) —Showing fine grain boundary sliding;  
 (b) —Showing whisker pull-out and good grain boundary bonding

in the initial strain rates within  $1.0 \times 10^{-2} \sim 1.0 \times 10^{-1}$  and at 893 ~ 903 K.

(2) The fracture surface of the composites, pulled at the optimum superplastic condition (i. e.  $3.3 \times 10^{-2} \text{ s}^{-1}$ , 903 K, 380%), shows extensive fine grain boundary sliding (grain size is about 1 ~ 4 μm), considerable whisker pull-out

and good grain boundary bonding.

(3) The DTA results show that for the composites the incipient melting temperature is 888.1 K, and the macroscopic melting temperature is 928.7 K.

(4) The composites can exhibit a good HSRS only in the narrow range of temperatures from 893 ~ 903 K. The primary mechanism for HSRS in the composites is fine grain boundary sliding and interfacial sliding accommodated primarily by an appropriately small amount of liquid phase.

## REFERENCES

- 1 Nieh T G, Henshall C A and Wadsworth J. *Script Metall Mat*, 1984, 18: 1405.
- 2 Imai T, Mabuchi M, Tozawa Y *et al.* *J Mat Sci Let*, 1990, 9: 225.
- 3 Mabuchi M, Imai T, Kubo K *et al.* *Materials Letters*, 1991, 11: 339.
- 4 Imai T, Esperance G L and Hong B D. *Script Metall Mat*, 1994, 31(3): 321.
- 5 Tochigi I, Imai T and Ai K. *Script Metall Mat*, 1995, 32(11): 1801.
- 6 Lim Suk-Won and Nishida Y. *Script Metall Mat*, 1995, 31(12): 1911.
- 7 Imai T, Kojima S and Jing D. *Mater Sci Eng*, 1997, A225: 184.
- 8 Imai T, Tochigi I, Ai K *et al.* *Script Metall Mat*, 1996, 34(10): 1627.
- 9 Mishra R S, Bieler T R and Mukherjee A K. *Acta Metall Mater*, 1997, 45(2): 561.
- 10 Mabuchi M and Higashi K. *Script Metall Mat*, 1996, 34(12): 1893.
- 11 Nutt S R and Carpenter R W. *Mater Sci Eng*, 1985, A75: 169.
- 12 Strangwood M, Hipsley C A and Lewandowski J J. *Script Metall Mat*, 1990, 24: 1483.
- 13 Zelin M G. *Acta Metall Mater*, 1997, 45(9): 3533.

(Edited by Huang Jinsong)