The Nature of Influence of Reinforcing Element Distribution on Superplastic Deformation Behavior of a Metal Matrix Composite

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Abstract

High temperature deformation behavior of the metal matrix composite 2009-15% SiCw with non-uniform and uniform distributions of whisker reinforcing elements has been studied by tension tests in the range of strain rates 10^{-5}-10^{2} s^{-1} at temperature 525^{0}C. Both states of the composite have shown superplastic-like behavior at this temperature. An increase of uniformity of reinforcing whiskers distribution in the composite microstructure results in a shift of the optimal strain rate of superplastic deformation toward higher strain rates and an increase of maximum value of coefficient of strain rate sensitivity *m*. Applied stress decreases. It is caused by the fact that the SiC whisker distribution influences the cooperative nature of grain boundary sliding (CGBS). In the composite with uniform distribution of the whiskers the spacing between surfaces of CGBS in units of average matrix grain size is slightly less than in the another state of the composite. In addition, the occurrence of grain boundary sliding was observed along a higher proportion of intergranular boundaries. Threshold stress was observed to decrease. The origin of high strain rate superplasticity (HSRS) in metal matrix composites (MMC) is discussed.

Introduction

It is known, that MMCs exhibit superplastic behavior at relatively high strain rates ($\dot{\varepsilon}\geq10^{-2}$ s^{-1}) [1]. In composites the optimal region of superplastic deformation (SPD) is from two to four orders of magnitude larger than that for monolithic aluminum alloys [1-5]. The origin of HSRS in MMCs is in present the subject of some debate. Nieh *et al.* [2] suggested a new rheological model in which semi-solid to be regarded as a Newtonian fluid. It was caused by the fact that the phenomenon of HSRS in MMCs has been observed at temperatures close to or above the matrix solidus temperature [1]. In addition, it was shown by use a special TEM technique that solute segregations of Mg and Cu at SiC/Al interfaces would reduce the incipient melting temperature and provide preferential melting of interfacial boundaries at temperatures of HSRS occurrence [1, 6]. However, the model [2] is in conflict with experimental results. If this model correct, the coefficient of strain rate sensitivity and elongation-to-failure may not depend from strain rate. In the same time the gap between HSRS of MMCs and SPD of monolithic alloys is not significant. There are not qualitative difference between superplastic behavior of the composite and their monolithic matrix alloys.

To solve this contradiction other models based on the important role of liquid phase in superplastic flow have been recently developed. *Mabuchi et al.* [3, 7] suggested that HSRS of the composites caused by relaxation of stress concentrations due to sliding along interfacial boundaries. However, a surface investigation [3] did not reveal any evidences of interfacial grain boundary sliding (GBS). GBS occurs along intergranular boundaries. In this connection an original model of SPD based on concept of isolated liquid phase at intergranular boundaries has been suggested by *Perevezentsev et* *al.* [8]. This theoretical model predicts the absence of significant difference between superplastic behaviors of MMCs and their monolithic matrix alloys. Summarized notice that the all models considered the presence of liquid phase in the body of MMC as necessity condition for HSRS, if correct, would support their contention that superplasticity in MMCs occurs by a unique deformation process. Therefore, perspectives of practical application of HSRS of MMCs in commercial scale would be restricted by service property degradation.

From other side, in case of creep it was proved that the deformation-controlling process in aluminum MMC is associated with deformation in aluminum matrix [9]. The same approach to HSRS of MMC has been developed in work [4]. It was shown [4] that the mechanisms of HSRS are similar to that for conventional superplasticity. The influence of reinforced elements on superplasticity was analyzed in terms of grain boundary sliding controlled by interface diffusion.

To investigate the role of ceramic reinforcement on superplastic deformation of MMCs, deformation behavior and structural evolution of two states of the 2009Al-15% SiC_W composite differed by distribution of the SiC_W has been analyzed.

Experimental materials and procedures

The composite 2009Al-15% SiC_W was produced via powder metallurgy from a standard aluminum alloy AA2009 (3.8% wt Cu, 1,3% wt Mg, 0.25% wt Si, Al is the balance) and 15% whiskers of SiC. The first state of the composite was obtained in the form of a mill processed 38 mm thick plate. The plate was subjected by cross-rolling (45% reduction at 530^{o}C and 45% reduction at 450^{o}C). It resulted in thick sheet reducing up to 1.5 mm. This is the second state of the composite. Two states of the composite are differed by uniformity of the SiC_W distribution (Fig. 1). In state I whiskers free areas are observed, while in state II they are almost absent The distribution of inter-whiskers spacing in the composite sheet is more uniform in comparison with the plate composite. Laboratory cross rolling resulted in approach of inter-whiskers spacing to matrix grain size.

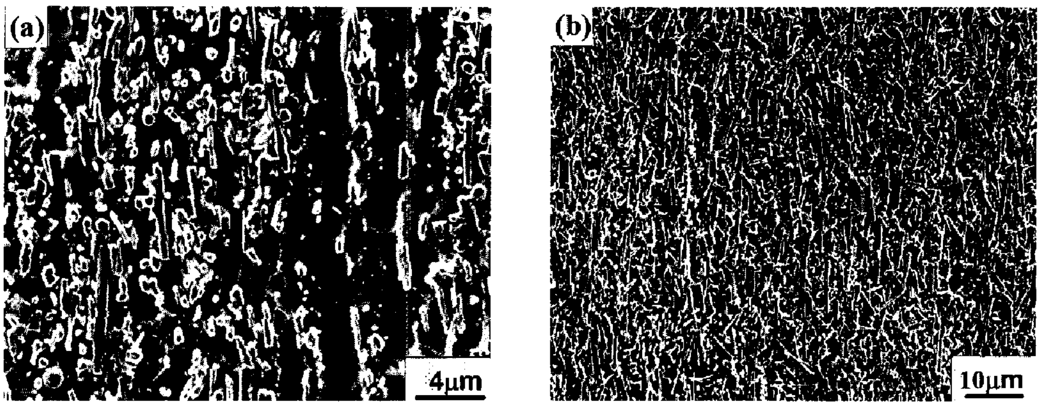

Figure 1. Microstructure of the composite 2009Al-15% SiC_W:
(a)- state I after mill processed plate; (b)- state II after cross-rolling sheet.

Tensile samples were machined from both the plate and the sheet in longitudinal direction with respect to the original extrusion direction. Samples with the gage section of 15x5x1.5 mm were tested in tension in air at temperatures 525^{0}C and at strain rates, from 10^{-5} to 10^{-1} s^{-1} using a Schenck PSA-100A universal testing machine. To evaluate mechanical properties the samples were tested up

to failure. The flow stress for each temperature-strain rate condition and for both states of the composite was determine as maximum value of stress.
For surface investigation the samples were prepolished by a diamond paste (0.5 μm); final polishing was performed using a 20% nitric acid solution in methanol at -30^{0}C and 15 V. Macroscopic scratches on some specimens were used to reveal a type of grain boundary sliding. The samples were deformed to the fixed strains at the optimal strain rates for superplasticity. Surface observation of deformed specimens was performed using a SEM JSM-840.

Results

Mechanical properties

Mechanical properties of the composite are similar to that reported in work [10]. Typical true stress-strain curves for both states of the composite are shown in Fig. 2. In state 1 all the curves exhibit an initial region of strain hardening at ε=5-20%. With decreasing strain rate the initial stage of plastic deformation becomes shorter. In stage 2, the flow stress gradually increases. Stress maximum is reached after the strain 30-40%. The peak strain does not significantly depend on strain rate. In stage 3, after the stress peak a gradual stress softening occurs until failure.
The sheet composite demonstrates slightly different type of σ-ε curves. The plastic flow of the composite is smooth and strain hardening rate is less in comparison with mill processed composite. After strains 70-110% the stable stage of plastic flow is reached.

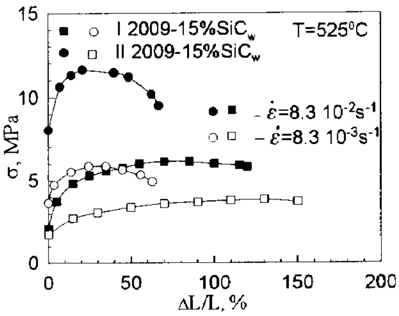

Figure 2. Stress-strain curves of the 2009Al-15% SiC_W composite.

Inspection shows that there is a sigmoidal relationship between the flow stress and the strain rate plotted on a double logarithmic scale for both states of the composite, divided into three distinct regions (Fig. 3). The flow stress is 1.5-2 times lower for the cross rolled material. The maximum value of the coefficient of strain rate sensitivity m=0.26 for the plate composite is less than that (m=0.32) for the cross rolled material. The maximum value of elongation-to-failure for cross rolled composite (190%) is 1.75 times greater than that for the mill processed composite (120%) (Fig. 4).

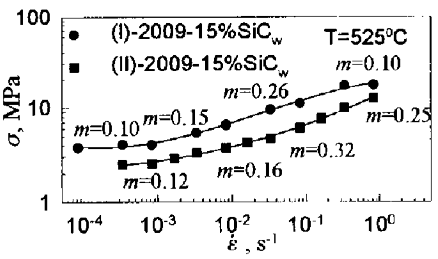

Figure 3. Strain rate dependence of flow stress.

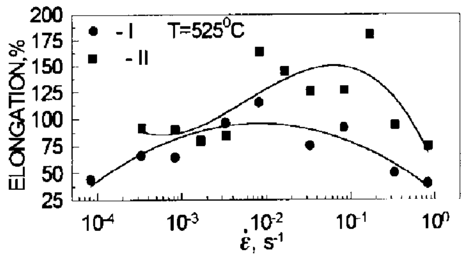

Figure 4. Strain rate dependence of elongation.

Analysis of deformation behavior of both states of the composite in terms of threshold stress is presented in Fig. 5. Threshold stress (σ_{th}) was obtained by extrapolation to zero strain rate of a straight line which the experimental data were plotted as $\dot{\varepsilon}^{1/n}$ against σ on a double linear scale at a single temperature at temperature 525^{0}C. The stress exponent of 3 yields the best linear fit between

$\dot{\varepsilon}^{1/n}$ and σ for both states of the composite. The datum points in this strain rate interval most closely fit a straight line whose extrapolation to zero gives the value σ_{th}. It is seen that threshold stress in the mill processed composite is significantly more than that in cross rolled composite.

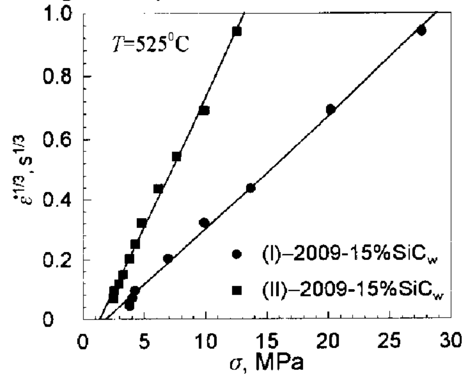

Figure 5. A plot of $\dot{\varepsilon}^{1/n}$ *vs* σ for the 2009Al-15% SiC_W composite at temperature T=525^{0}C.

Surface microstructural observation

Mill processed composite. It has been established that surface metallohraphic features depend on strain and strain rate. Micrographs of the relief evolution in region 2 are shown in Fig. 6. Investigations of the specimen surface after deformation revealed direct evidences for CGBS operation [11-13] in aluminum matrix. In both states of the composite the interfacial sliding is not observed while there is evidence of extensive sliding along intergranular boundaries (Fig. 6, 7).

After strain ε=10% thin longitudinal deformation bands appear along intergranular boundaries at a distance equal to two units of average initial matrix grain size (Fig. 6a). Observation of these bands is associated with the shifting of matrix grain groups as a unit along common grain boundary surfaces [11-13]. In macroscopic scale the bands form stringers, which traverse the specimens from one edge to the other. These stringers are discontinuous. In some cases the barriers to band propagation are the SiC_W. All the bands tend to align over and are oriented at an angle of 60° to the tension axis. Transverse bands have not been observed.

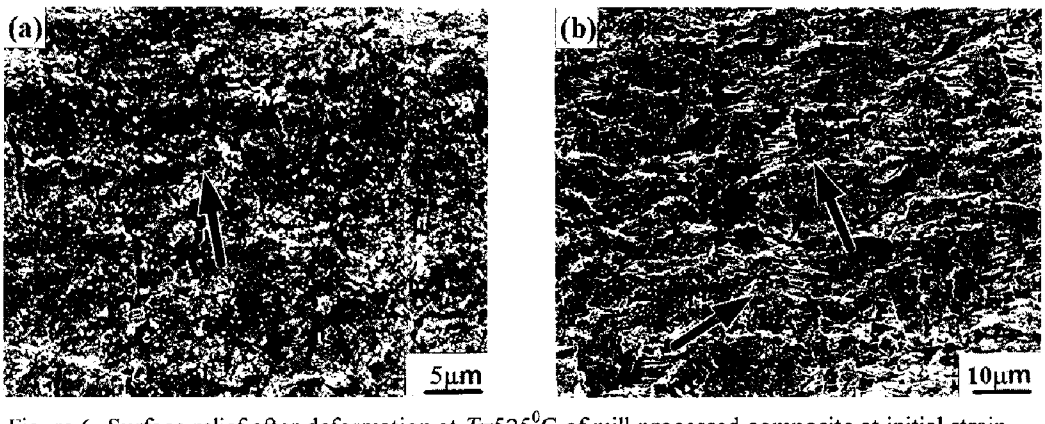

Figure 6. Surface relief after deformation at T=525^{0}C of mill processed composite at initial strain rate of $\dot{\varepsilon}$=8.3·10^{-2} s^{-1}; (a) ε=10%, (b) ε=100%. Arrows indicated directions of the stringers. The tensile axis is horizontal

Further deformation up to strain ε=100% leads to a sharp increase in deformation band broadening (Fig. 6b). At the same time the spacing between surfaces of CGBS remains virtually unchanged. Unlike monolithic alloys [12, 13] propagation of deformation bands is retarded by whiskers and it becomes discontinuous in manner. The most important feature of the deformation relief after large strain is the formation of narrow transverse deformation bands. Their width is smaller than the width of longitudinal bands by a factor of 5. Connection of secondary transverse bands with primary longitudinal bands leads to formation of a band net which segments the entire matrix volume. The longitudinal band stringers align at an angle of about 60° to the tension axis.

Cross rolled composite. In cross rolled composite after small strain (ε=10%) thin deformation bands combine into the stringers in macroscopic scale (Fig. 7a). These stringers are rather continuous. The average distance between stringers is equal to about two units of average initial matrix grain size and is slightly more than that in mill processed composite (Fig. 6a). Notice that broadening of deformation bands in sheet composite is more than that in the mill processed composite.

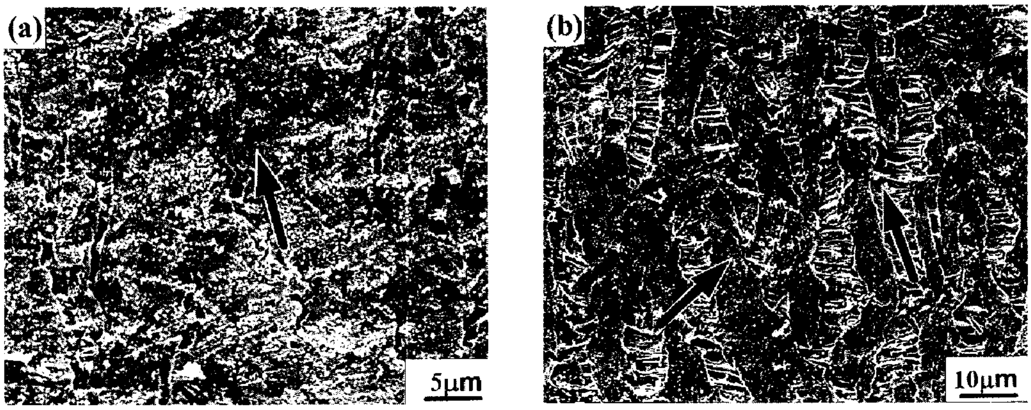

Figure 7. Surface relief after deformation at T=525°C of cross rolled composite at initial strain rate of $\dot{\varepsilon}=8.3\cdot10^{-2}$ s^{-1}; (a) ε=10%, (b) ε=100%. Arrows indicated directions of the stringers. The tensile axis is horizontal

Further deformation up to ε=100% results in decrease of spacing between the stringers and minor increase of deformation band broadening (Fig. 7b). The average distance between CGBS surfaces is the same than that in the mill processed composite. However, the character of CGBS in the sheet composite differs from that in the plate composite. Operation of single system of cooperative grain boundary sliding was revealed even after the great strain. The continuous straight stringers locate in parallel with one another. The evidences of operation of secondary system of CGBS were not found.

Discussion

It was shown in the prior work [10] that no segregation or liquid phase formation on interfacial boundaries was observed at the testing temperature 525°C. From presented results it is seen that the rate-controlling deformation process in the 2009Al-15%SiC_W composite is associated with deformation in the aluminum matrix. Consequently, this implies that the role of liquid phase is not important for the origin of HSRS in the composite at this temperature. The consistent feature of presented experiments of surface microstructural observation is that sliding along intergranular boundaries is the dominant flow mechanism. There are strong grounds to believe that the contribution from GBS in aluminum matrix to the total elongation of the composite is as high as in monolithic aluminum alloys (about 50%-70% in region 2) [5]. This mechanism of plastic deformation is responsible for features of deformation behavior of two states of the composite and advent of HSRS in MMCs.

CGBS in MMCs is much more uniform in comparison with monolithic aluminum alloys. The uniformity of grain boundary sliding is characterized by the size of sliding grain groups. The spacing between surfaces of CGBS is an important parameter of superplastic deformation [12]. The size of grain groups effects the optimal strain rate of superplasticity, the total elongation and the maximum value of the strain rate sensitivity m [12, 13]. Reduction of the grain group size results in a shift of the optimal region of superplastic deformation toward higher strain rates [12]. It is known [11-13] that in single-phase monolithic alloys the spacing in units of average grain size is from 5 to 8 at a small strain and from 2 to 5 at a high strain. In the composite the spacing between surfaces of CGBS is much less than in monolithic alloys and does not exceed two average matrix grain sizes. Moreover, the influence of strain on the spacing is insignificant. Therefore, introduction of SiC reinforcements provides the great uniformity of superplastic deformation. As a result, the region 2 of superplasticity in MMCs shifts toward higher strain rates. Thus, the homogeneous of CGBS is dominant factors determining the advent of HSRS in MMCs at pre-melting temperatures.

The influence of cross rolling on superplastic behavior of the composite is associated with the homogeneous of CBS. In cross rolled composite the distribution of the SiC whiskers is more uniform than that in mill processed composite. As a result, grain boundary sliding occurs under the traction of the maximum shear stress, and operation of a preferential system of CGBS is observed at all stages of plastic flow [13]. The grain groups have a right-angle shape. This provides the low flow stress and the relatively high value of the coefficient of strain rate sensitivity m in the sheet composite. In mill processed composite the strain increase leads to operation of multiple CGBS. The shape of grain groups becomes rather equiaxed. The gap between σ-ε curves in two states of the composite is caused by the difference in character of CGBS. At present, no model can explain the observed gap between threshold stresses for two states of the composite satisfactory.

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