SUPERPLASTICITY OF THE METAL MATRIX COMPOSITE PM2014 AL-20%Al_2O_3

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ABSTRACT.

The hot deformation behavior of the aluminum alloy metal matrix composite PM2014-20% Al_2O_{3p}, produced via powder metallurgy, has been studied by tension tests in the range of strain rates $\dot{\varepsilon}=10^{-4}$-1.1 s^{-1} at temperature T=500°C. The composite has shown superplastic behavior at these temperatures. The coefficient of strain rate sensitivity "m" was computed. Optical metallographic studies and TEM investigations were performed. It has been shown that superplastic deformation leads to formation of band microstructure. The reasons of such behavior of the composite during hot deformation are discussed.

INTRODUCTION.

Discontinuously reinforced aluminum alloy metal matrix composites (**MMCs**) exhibit a unique combination of high modules and strength at room temperature. Recent results have displayed the effect of high strain rate superplasticity (**HSRS**) in numerous aluminum alloys, **MMCs** produced via both powder metallurgy and squeeze-casting [1-6]. This property of the composites allows to use extrusion and forging technologies developed for conventional aluminum alloys. It is known, that composites exhibit superplastic behavior at temperatures above the solidus point and at relatively high strain rates ($\dot{\varepsilon}>10^{-2}$ s^{-1}). In the **MMCs** the optimal superplastic strain rate range is from two to four orders of magnitude larger than that for conventional aluminum alloys [1, 6-10]. A satisfactory explanation of this phenomenon does not exist. It is caused by the fact that the main attention was given to mechanical properties whereas the structural changes during superplastic deformation were insufficiently investigated. The aim of the present paper is to study the effect of plastic deformation condition on mechanical properties and microstructure of the **MMC.**

MATERIAL AND EXPERIMENTAL PROCEDURES.

The composite PM2014-20% Al_2O_{3p} was produced via powder metallurgy from a standard aluminum alloy AA2014 (4.55% Cu, 0.6% Mg, 0.78% Mn, 0.86% Si, 0.06%Fe) and 20% particles of alumina. Tensile specimens, 10 mm in gauge length and 4mm in width, were machined from the extruded bar ∅15mm, with the gauge length parallel to the extrusion direction. Tension tests were carried out on a "Instron 1185" universal testing machine at temperature t=500°C and strain rate $\dot{\varepsilon}=10^{-4}$-1.1 s^{-1}. The values of strain rate sensitivity $m=\partial(\lg\sigma)/\partial(\lg\dot{\varepsilon})$ were measured by the strain rate jump test [11]. Metallographic analysis of deformed specimens was performed by means of an optical microscope "Neophot-32", a scanning electron microscope "Jeol-840" and a structural analyzer "Epiquant". Specimens for optical microscopy were prepared by mechanical polishing and by etching in standard Keller solution. The fine structure was examined in "Jeol-2000EX" electron microscope on thin foils jet polished to perforation using a 20% nitric acid solution in methanol at -30°C and 15 V. Dislocation density was determined by number of dislocations intersections from the foil surface [13].

RESULTS.

Mechanical properties. Typical true stress-strain curves for the composite are shown in fig.1. The curves σ-ε for all temperature-strain rate conditions of deformation indicate three stages of plastic flow. The sharp strain hardening takes place at the early stage of deformation at ε=2-10% . At stage 2, the flow stress continues to increase, but more slowly than at stage 1. At stage 3, after a stress peak the flow stress slowly decreases. Stress maximum is reached after large strains (ε=70-110%). The largest peak strain was obtained at $\dot{\varepsilon}=8.3\cdot10^{-2}$ s^{-1}. An increase or decrease of the strain rate results in a shift of the stress peak to a range of small strains.

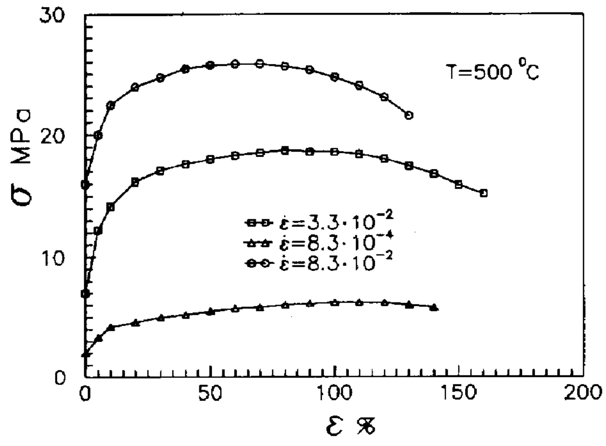

Fig. 1: Stress-strain curves for PM2014-20% Al_2O_3p.

Inspection shows that there is a sigmoidal relationship between the flow stress and the strain rate plotted on a double logarithmic scale [17]. The deformation behavior is divided into three distinct regions. In the first and third regions an increase in the strain rate leads to a slow increase of a flow stress level whereas in the second region the σ-ε dependence is very strong. The values of coefficient "m" are different in these regions. In the second region at $\dot{\varepsilon}=3.3\cdot10^{-2}$ s^{-1} the value of m is maximum (m=0.45). In the first and third regions the values of m are about 0.20 and 0.25 , respectively.

Total elongation has maximum (δ=210%) at t=500°C and $\dot{\varepsilon}=2.8\cdot10^{-2}$ s^{-1} [17]. A strain rate change results in smooth reduction of total elongation. In general, large elongation "δ" was observed in the temperature-strain rate region where high values of "m" coefficient were obtained. However, a maximum of strain rate sensitivity coefficient shifts toward higher strain rates relative to a maximum of total elongation.

Microstructure. Optical microstructural investigation has shown that in the pre-existing state the composite has an anisotropy distribution of alumina particles (fig.2a). Reinforced particles are extended in the direction of extrusion. Their average length is 6-8 μm. In transverse direction their form is circular and their average size is about Ø2 μm. The distribution particle is not completely uniform. Clusters of reinforced elements are observed in the composite. The matrix microstructure also has a small anisotropy: the grains have some elongation in the prior extrusion direction. In trasverse direction it has a circular form and their average size is 2.5 μm. Prolonged annealing at temperatures near that of solidus does not cause their size and form changes.

Hot plastic deformation changes a microstructure of the material. The character of microstructure transformation depends on a strain rate. Unusual evolution of microstructure is observed in the optimal temperature-rate interval of superplastic deformation (fig.2b-d). Deformation bands are formed along tension direction after small strains ε=2-3%. The length of deformation bands reaches several tens of microns and their width is determined by the distribution of particles at the site of specimen and constitutes several microns. With increasing a strain up to ε=30% the broad of deformation bands grows and their volume fraction increases. After this strain two structural components can be distinguished in the composite microstructure. The first component is a banded structure. It occupies approximately 30% of the material volume and is observed at sites with a lower density of reinforced particles. The second component is comprised of circular grains of about several μm in size. It is observed at sites of reinforced element clusters. The subsequent strain up to ε=50% results in a decrease of the first structural component volume fraction and provides an increase of the fine grains volume fraction. After ε=100% the banded structure disappears (fig.2c).

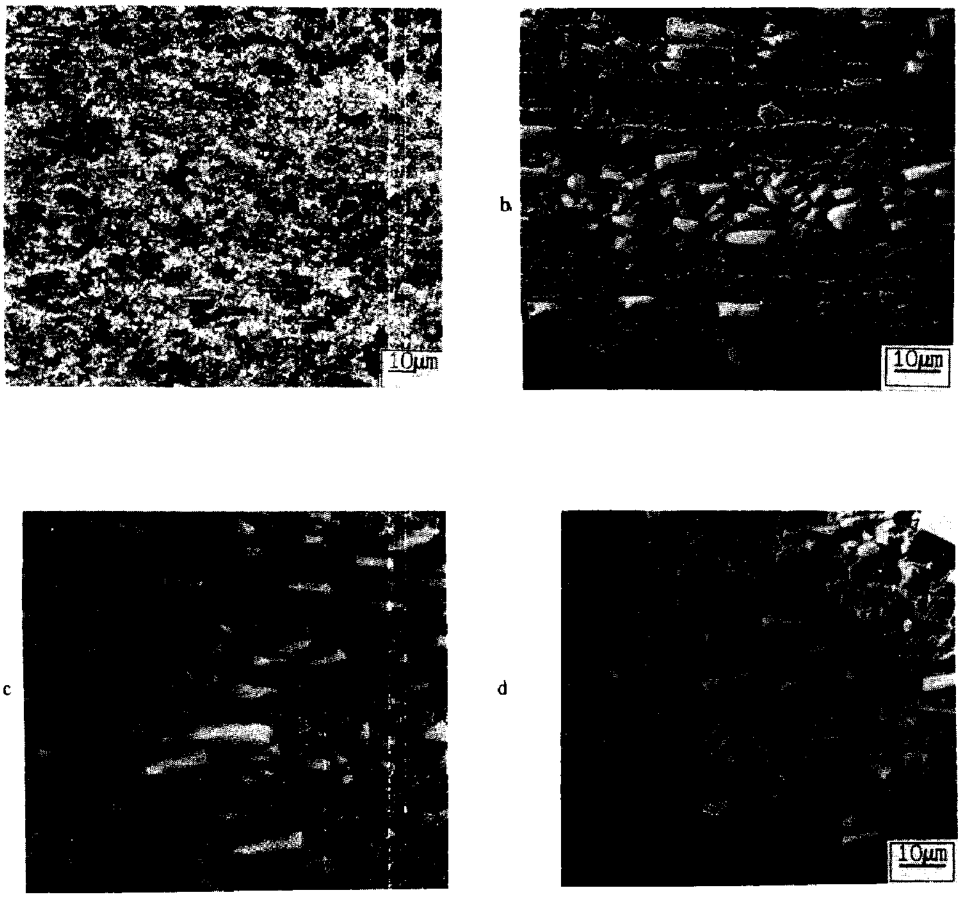

Fig. 2 Microstructure of the composite:
a) as-received state; b) t=500°C, $\varepsilon=3.3\cdot10^{-2}\ s^{-1}$, ε=30%;c) t=500°C, $\varepsilon=3.3\cdot10^{-2}\ s^{-1}$, ε=100%;
d) t=500°C, $\varepsilon=1.6\cdot10^{-4}\ s^{-1}$, ε=100%.

In the first region at the interval of low strain rates the structural evalution is different (fig.2d). The formation of banded structure at early stages of plastic flow is almost not revealed. An intensive growth of initial grains has been observed and their size has become approximately similar to that of deformation bands. With increasing the strain the initial grains grow and absorb the microband structure. At ε=30% a formation of homogeneous structure is observed. The deformation at the stable stage of plastic flow leads to a further growth of grains.

The increase of the strain rate above the superplasticity optimum firstly accelerates the formation of recrystallized grains at places of prior deformation bands and then suppresses the formation of deformation bands at the early stage of plastic flow. At strain rates higher than $\varepsilon=10^{-1}\ s^{-1}$ a refinement of initial grains has been observed. The formed grains are elongated in the tension direction.

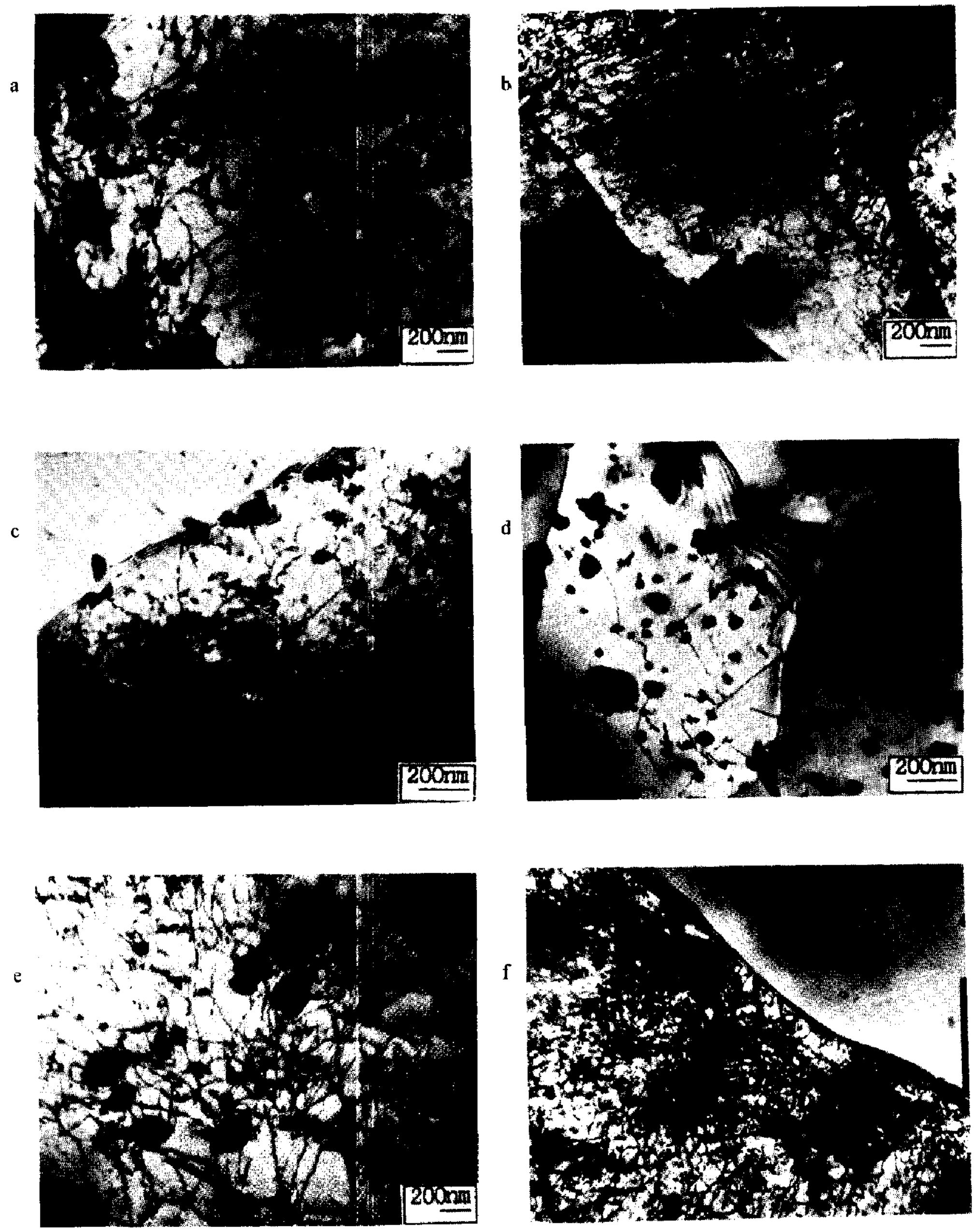

Fig. 3: Fine structure of the composite ($t=500°C$)
a) $\varepsilon=3.3\cdot10^{-2}\ s^{-1}$, $\varepsilon=15\%$; b) $\varepsilon=3.3\cdot10^{-2}\ s^{-1}$, $\varepsilon=30\%$; c,d) $\varepsilon=3.3\cdot10^{-2}\ s^{-1}$ $\varepsilon=50\%$,
e) $\varepsilon=1.6\cdot10^{-4}\ s^{-1}\varepsilon=100\%$, f) $\varepsilon=1.1\cdot s^{-1}$, $\varepsilon=100\%$.

Fine structure. Low angle boundaries are observed very seldom in the initial structure of the material. Deformation bands have not been revealed. Grain boundaries show distinct extinction contract. There are grain boundaries dislocations (**GBD**) in some grain boundaries at the same time. The density of lattice dislocations in the annealed as-received material is not so high $\rho=3\pm1.5\cdot10^9$ cm^2 . The dislocations are distributed uniformly over the matrix volume.

Deformation in the optimal superplastic interval leads to a decrease of lattice dislocation density in initial grains and increases it in formed deformation bands (fig.3a-f). After strain $\varepsilon=15\%$ the dislocation density in the latter is $\rho=8\cdot10^9$ cm^2(fig.3a). The increase of the strain up to $\varepsilon=30\%$ leads to the formation of subgrain boundaries in deformation bands (fig.3b). After $\varepsilon=50\%$ two types of grains with different dislocation density are observed in the composite. In some grains $\rho=8.2\cdot10^8$ cm^2 that is lower than in the initial state of the composite (fig.3d). A higher dislocation density is observed only near some interfaces Al/Al_2O_3 (fig.3c). In other grains the density is considerably higher ($\rho=2.4\pm1.3\cdot10^{10}$ cm^2). Further deformation causes a decrease of dislocation density in this type of grains.

In case of decreasing the strain rate the dislocation density after $\varepsilon=50\%$ firstly grows and then decreases. At $\dot{\varepsilon}=8.3\cdot10^{-4}$ s^{-1} the plastic deformation causes a rise of dislocations density to $\rho=8.5\pm3\cdot10^{10}$ cm^2. The decrease of the strain rate to $\dot{\varepsilon}=1.6\cdot10^{-4}$ s^{-1} reduces the density of dislocations to $\rho=9\cdot10^8$ cm^2 (fig.3e). The increase of the strain rate above an optimal interval of superplastic deformation (**SPD**) always causes a growth of lattice dislocation density. Thus, at the strain rate $\dot{\varepsilon}=1.2\cdot10^0$s^{-1} the dislocation density after $\varepsilon=100\%$ constitutes $\rho=3\cdot10^{10}$cm^2 (fig.3f). One should note that within the whole strain rate interval of deformation grain boundary dislocations have been observed.

DISCUSSION.

As seen from the present results, a shape of σ-ε curves, microstructure and substructure evolution during deformation and relatively high rates of an optimal interval of **SPD** are not typical for conventional superplasticity [7,9,10]. At the same time, the influence of a strain rate on the flow stress, values of the strain rate sensitivity coefficient "m" and the total elongation δ testify exactly that **SPD** at t=500°C and $\dot{\varepsilon}=8.3\cdot10^{-4}$-$8.3\cdot10^{-2}$ s^{-1} takes place [7,9,10]. This contradiction can be explained in terms of cooperation grain boundary sliding (**GBS**) and migration [11,12,14] taking into account the peculiarities of the aluminum alloy matrix.

The fine oxide particles present in the aluminum matrix and other impurities, as a result of composite manufacturing via powder metallurgy [15], cause the occurrence of threshold stresses in the composite (their value constitutes 2.8 MPa in this composite), and considerably reduce the ability of grain boundaries to absorb dislocations. The presence of **GBD** even after two hours annealing at near melting temperature attests to the latter. These factors influence on the main operative mechanism of **SPD** - **GBS**.

In region 1 the climb controlled dislocation sliding is main mechanism of deformation [7]. At strain rate $\dot{\varepsilon}\geq8.3\cdot10^{-4}$ s^{-1} grain boundaries are unable to absorb all lattice dislocations within the matrix grains. Due to that the dislocation density increases with rising the strain and this causes strain hardening at $\varepsilon\leq50\%$. At $\dot{\varepsilon}\leq8.3\cdot10^{-4}$ s^{-1} the growth of dislocation density inside grains is not observed and plastic flow of the composite attends a stable stage after small strain.

In region 2 flow stress becomes higher than threshold stress for movement of **GBD** [8]. **GBS** becomes main mechanism of plastic deformation. At the places of reinforced element clusters **GBS** occurs as mutual displacement of single grains according to a classic scheme [7] as a homogeneous process. Intensive **GBS** accelerates the absorption of lattice dislocations by grain boundaries [7]. The density of dislocations reduces. The exception is the regions being at some distance from the intergranular boundaries Al/Al and adjacent to the interfaces Al/Al_2O_3. **GBS** does not develop along interfacial boundaries and this results in the formation of dislocation pile-ups here.

At the sites with a lower concentration of reinforced particles **GBS** develops as a planar mode of deformation and results in the formation of bands of intensive deformation. This structure formed in the composite after small strains is similar to that described early for the Zn-22% Al alloy [11]. Deformation localizes at the sites free from reinforced particles. The cooperative **GBS** becomes the most important mechanism of **SPD** at the early stage of plastic flow. It is known [14], that a cooperative manner of **GBS** causes the cooperative grain boundary migration (**GBM**). The combining of the processes of **GBS** and **GBM** leads to a growth of band broad [16]. Under these conditions intensive **GBM** inside deformation bands results in rapid formation of coarse elongated grains after relatively small strains [16]. The development of **GBS** on such grains ceases. The change of the main mechanism of deformation occurs in deformation bands. Intragranular dislocation slip becomes the main deformation mechanism. Consequently, the contribution of cooperative **GBS** to the total deformation reduces and that of homogeneous **GBS** increases. In the coarse grains, formed at the site of prior deformation bands, dynamic recrystallization occurs. This leads to the formation of a fine-grained structure after high strains. During plastic flow rearrangement and alignment of reinforced elements take place, providing their homogeneous distribution over the composite volume. Evidently, this explains the fact that homogeneous **GBS** mainly contributes to the total deformation at the latter stage of plastic flow.

Thus, the strain dependence of contributions of cooperative **GBS** and homogeneous **GBS** to the total deformation is the reason of unusual high temperature deformation behavior of the PM2014-20% Al_2O_{3p} composite.

CONCLUSION.

1. The PM2014-20% Al_2O_{3p} alloy shows superplastic behavior at high homologous temperature and high strain rates. The shape of σ-ε curves, structural and substructure evolution during deformation is not typical for conventional superplasticity. They are caused by the action of a cooperative manner of grain boundary sliding.

2. Superplastic deformation leads to the decrease of dislocation density in some grains. Simultaneously, at the sites with a lower concentration of reinforced particles the deformation bands are formed with dislocation density being higher by two orders. With increasing a strain the development of dynamic recrystallization in the latter results in formation of new grains containing a small number of dislocations.

3.The strain rate vs dislocation density dependence of the composite at t=500°C after large strain has a minimum at the optimal rate of superplasticity.

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