Proceedings of the 12th International Symposium on Physics of Materials, Prague, September 4–8, 2011

Mechanical Properties of Aluminum–Matrix–Nanoparticle Composites

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Aluminum-matrix-nanoparticle composites were produced by ball milling of micro scale aluminum powder with different additions of 500 nm mean diameter SiC particles in air atmosphere with subsequent consolidation by hot extrusion. The material was investigated in this condition by tensile testing, hot tensile testing, density, hardness, and amplitude dependent damping measurements. The amplitude dependent damping of the material was investigated after slow furnace heating and cooling as after quenching into water to the room temperature. The results of the tensile and hot tensile testing show that the addition of SiC particles lead to a significant increase of tensile strength, but the remaining porosity increased, too. The results obtained for the amplitude dependent damping can be attributed to cracks present in the consolidated material or cracks produced by thermal stresses.

PACS: 61.82.Rx, 62.40.+i, 81.05.−t, 81.05.Ni, 81.07.Wx, 81.20.Ev, 81.20.Hy, 81.40.Ef, 81.70.−q, 81.70.Bt

1. Introduction

Dispersion hardening is a well known method to strengthen metals in a broad temperature range [1]. For this the intermediate free distance between the particles must be as small as possible. This can be achieved by the distribution of nanoscale particles homogenous even for small volume fractions in the order of 1 to 10%. For this publication the homogeneous distribution was attempted by ball milling of aluminum powder with nanoscale SiC powder and subsequent hot extrusion.

In many cases the degree of consolidation cannot only be characterized by the porosity remaining in the consolidated material. Cracks or microcracks must be considered, too. Therefore the mechanical properties of aluminum composites with nanoscale ceramic particles produced by ball milling and hot extrusion were investigated with tensile tests at room temperature and at the 300 °C, hardness measurements, density measurements and amplitude depending damping measurements.

2. Experimental

Aluminum-matrix composites were prepared by ball milling of water atomized aluminum powder, with 25 μm and 45 μm mean diameter fractions produced by TLS Technik Spezialpulver (Bitterfeld, Germany), and 200 nm–1 μm (500 nm mean diameter) SiC Powder (Alfa Aesar GmbH, Karlsruhe, Germany), mixed with 0.5 g of stearic acid, in a planetary ball mill PM400 (Retsch, Haan, Germany) under air atmosphere. The two aluminum powder fractions, each of 50 vol.%, were mixed together with silicon carbide and stearic acid to achieve the secondary powder convenient for the following consolidation. A corundum-coated steel milling cup with a volume of 500 ml was filled with the powder mixture and 7 mm diameter and 13 mm diameter light corundum balls. The ratio of the mass of powder to ball 7 mm to ball 13 mm was 1:1.5:1.5, respectively [2]. The powder volume in the milling cup being about 20% of the whole volume was milled for 6 h at 150–200 rpm. Oxidation, plastic deformation, and fracture of the aluminum powder particles took place during the milling and led to a fine Al_2O_3 and SiC nanoparticle distribution in the plastically deformed aluminum particles. Thus a non-agglomerated secondary metal matrix composite (MMC) powder with flake-like particles of about 20 μm diameter and 5 μm thickness was produced.

After ball milling the secondary MMC powder was filled into aluminum capsules of 70 mm diameter, 2 mm wall thickness, and about 220 mm height, closed at one end with a circular plate of 5 mm thickness. The powder was multistage cold pre-pressed with a pressure of 65 MPa. The capsules were repeatedly filled to about one half of the remaining height after successively pressing up to 210 mm height of the pressed powder. Afterwards the capsule was closed by another circular plate of 5 mm thickness with a small hole of 2 mm diameter in the

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middle using fitting threads in the capsules and plates. The small hole acted as gas outlet during hot extrusion.

The pre-compacted samples were hot extruded to rods of 20 mm diameter (1:3.5 extrusion ratio) using a horizontal 630 t hot extrusion press. The extrusion temperature and velocity were 450 °C and 2 mm/s, respectively [3]. The names of the consolidated materials with various additions of SiC and Al_2O_3 particles are listed in Table I. It has to be stressed that the given volume fractions are additions or nominal values. The volume fraction of alumina particles in the secondary powder and in the consolidated rods is higher due to the surface oxides on the used aluminum powder and the oxidation of aluminum during milling.

TABLE I

Names of investigated samples for various additions of SiC and Al_2O_3 dispersoids.

Name of sample	Addition of dispersoids in vol.%	
	SiC	Al_2O_3
Al–HE	0	0
Al–4%SiC	4	0.15
Al–6%SiC	6	0
Al–8%SiC	8	0.7
Al–10%SiC	10	1.4
Al–13%SiC	13	0

Samples for tensile and hot tensile testing with 5 mm diameter and 56 mm long were machined out of the rods. For density measurements, cylindrical samples (60 mm long) were cut from the rods. Bending samples for damping measurements, 3 mm thick, 10 mm wide, and 80 mm long with cylindrical strengthening of 10 mm diameter and 30 mm length on one side were machined out of the rods, too.

The strain amplitude dependent damping was determined by measuring the logarithmic decrement of freely decaying vibrations of the bending beams clamped at the strengthened end. The specimens were excited into resonance by a permanent magnet fixed at the free end of the bending beam and a sinusoidal alternating magnetic field. The resonance frequency ranged from 30 to 40 Hz. Damping was measured as the logarithmic decrement δ of free decaying vibrations [4, 5].

3. Results

The results of the tensile tests at room temperature for various samples are shown in Fig. 1. It is obvious that the yield and tensile strength increase and the ductility decreases with increasing volume fraction of SiC particles (see Table I).

Figure 2 shows the results of hot tensile tests carried out at 300 °C with different true strain rates. The true strain rates $\mathrm{d}\varepsilon/\mathrm{d}t = \mathrm{d}l/(l(t)\mathrm{d}t)$ are plotted versus the true stress $\sigma = F(t)/q(t)$ in log–log plot with

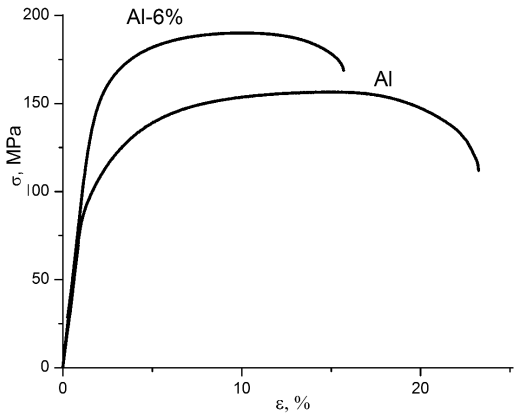

Fig. 1. Engineering stress–strain curves measured at room temperature for hot extruded ball milled aluminum powder and material with 6 vol.% SiC addition. $\dot{\varepsilon} = 3.3 \times 10^{-4}$ min^{-1}.

$l(t)$, $F(t)$, and $q(t)$ being the length, force, and cross-section of the sample, respectively, at time t where the stress is measured. The high stress sensitivity exponents $n = \mathrm{d}\ln(\mathrm{d}\varepsilon/\mathrm{d}t)/\mathrm{d}\ln(\sigma)$ ranging from about 14 to 20 indicate the power law breakdown for the high temperature deformation (creep) mechanism. It can be seen that the stress level increases substantially. Both figures (Fig. 1 and Fig. 2) indicate dispersion hardening by SiC particles.

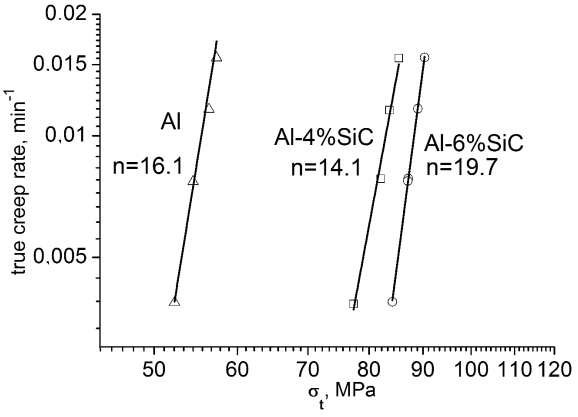

Fig. 2. True stress dependence of true creep rate for tensile tests carried out at 300 °C for samples with different SiC volume percentage in the exponent in a log–log scale. n is the stress sensitivity exponent.

The dependence of the volume fraction of SiC particles on the yield strength at constant strain rate of $\dot{\varepsilon}_{\mathrm{RT}} = 3.3 \times 10^{-4}$ min^{-1} and $\dot{\varepsilon}_{300^\circ\mathrm{C}} = 7.7 \times 10^{-3}$ min^{-1} is presented in Fig. 3. A nearly linear increase of yield strength is observed, when the yield strength is plotted versus the added volume fraction of SiC. Moreover, the slope of the straight line of the room temperature measurements is only a little higher compared to the one of the 300 °C measurements.

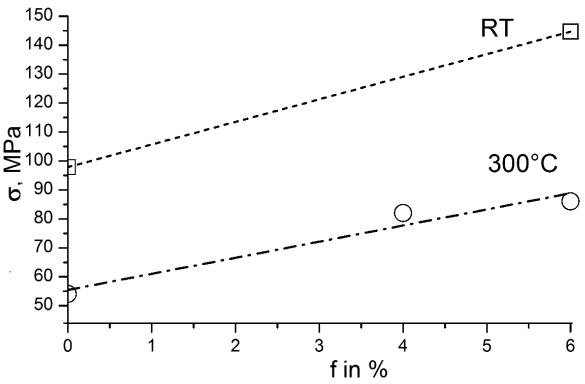

Fig. 3. Yield strength at room temperature RT and at 300 °C versus volume fraction of SiC particles with mean diameter of about 500 nm at $\dot{\varepsilon}_{\mathrm{RT}} = 3.3 \times 10^{-4}$ min^{-1} and $\dot{\varepsilon}_{300°\mathrm{C}} = 7.7 \times 10^{-3}$ min^{-1}.

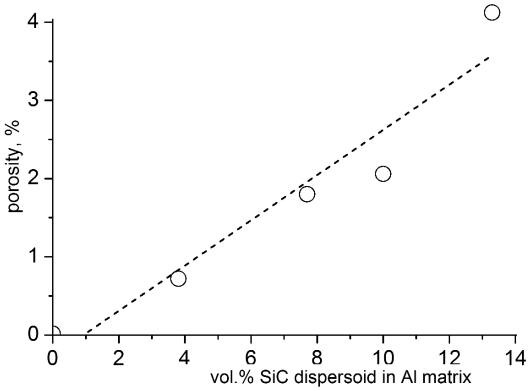

Fig. 4. The remaining porosity after hot extrusion at 450 °C vs. volume percentage various additions of SiC--dispersoids.

With increasing addition of SiC dispersoids the remaining porosity of the material after hot extrusion increases, too. This can be observed in Fig. 4 indicating a linear dependence or even the proportionality of the two items.

The additional consolidation or sintering behavior after successive isochronal heat treatment for 1 h at increasing temperatures was studied by density and hardness measurements shown in Fig. 5. Both the hardness and density increase with increase of the annealing temperature, indicating the preceding compaction of the sample. The linear regression of density and hardness versus the annealing temperature lead to the straight lines in Fig. 5 fitting the results. The relative slope of the hardness is found to be $\mathrm{dHV/HV(RT)}\,\mathrm{d}T = 5.3 \times 10^{-5}$ K^{-1} while the relative slope of the density yields $\mathrm{d}\rho/\rho(\mathrm{RT})\,\mathrm{d}T = 4.8 \times 10^{-6}$ K^{-1}.

Therefore, the amplitude dependent damping of the samples was measured for a material without SiC additions in order to detect the damping contribution of microcracks that cannot be observed by conventional tech-

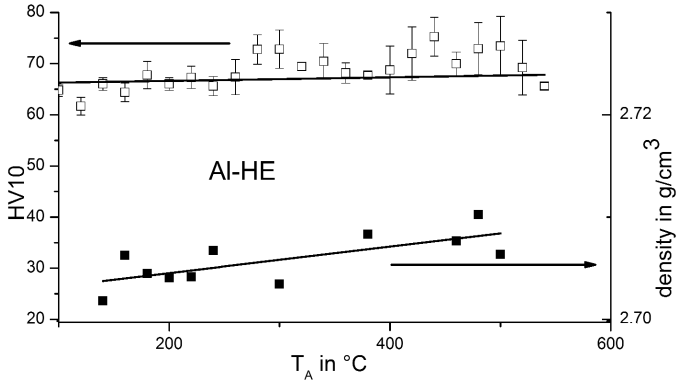

Fig. 5. Hardness and density versus annealing temperature for successive isochronal annealing for 1 h and increase of the annealing temperature. The straight lines represent linear regressions of the measured points. The relative slope of hardness $\mathrm{dHV/HV(RT)}\,\mathrm{d}T = 5.3 \times 10^{-5}$ K^{-1}. The relative slope of the density $\mathrm{d}\rho/\rho(\mathrm{RT})\,\mathrm{d}T = 4.8 \times 10^{-6}$ K^{-1}.

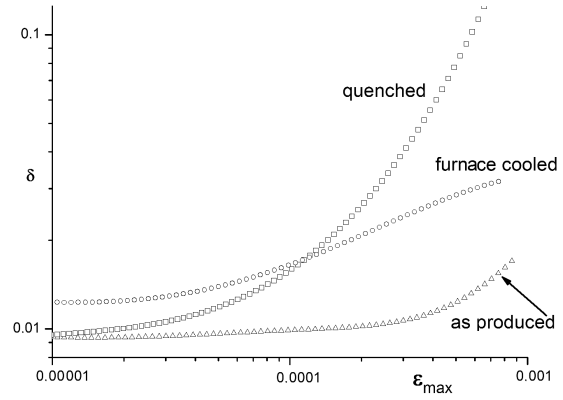

Fig. 6. The logarithmic decrement of the freely decaying bending beam vibration versus maximum strain amplitude in the bending beam in a double logarithmic plot. The tested material was ball milled and hot extruded aluminum powder in as produced and heat treated (400 °C, 24 h) conditions whereas one sample was water quenched and the other slowly (furnace) cooled. The resonant frequencies were 30.0 Hz for quenched, 33.4 Hz for furnace cooled, and 32.6 Hz for the as-hot-extruded material.

TABLE II

Mechanical properties of powder metallurgical produced aluminum and aluminum with 6 vol.% SiC addition.

Mechanical property	Name of sample	
	Al	Al–6%SiC
yield strength [MPa]	93	145
ultimate strength [MPa]	157	190
ultimate elongation [%]	14	7.8
total elongation [%]	22	13

niques like penetration or radiation testing. The results are plotted in Fig. 6 as the logarithmic decrement (damping) versus the maximum strain amplitude of vibration bending beams. The high strain independent damping of the as hot extruded material compared to the low damping of conventionally produced aluminum indicates a broad distribution of small cracks present in the investigated sample. Quenching the bending beams from 400 °C into water of room temperature increases damping at higher strain amplitudes substantially, which can be attributed to the creation of cracks due to the thermal stresses during quenching. Slow furnace cooling results into different damping behavior. The damping at higher strain does not increase very much due to the heat treatment but increases a little with the increase of the strain amplitude beginning at unusually small strain amplitudes. This indicates that not all cracks have been closed during the heat treatment, but on the contrary they have been increased a little, although the density increased, too.

4. Discussion

The homogeneous distribution of dispersoids in a metal matrix is a necessary condition for optimal material strengthening by dispersoids. The dispersion strengthening with non-shearable particles up to 10 vol.% can be explained with the Orowan mechanism, resulting in an additional shear strain of $\tau_0 = \alpha G b/l$, where G is shear modulus, b is the Burgers vector, l is the free distance between two dispersoids and α is a proportionality factor depending on the arrangement of the neighboring particles. For homogeneously distributed dispersoids the mean distance between two of them is about $l \approx d/\sqrt{f}$, where d is the mean diameter of the particles and f is their volume fraction of the dispersoids. Therefore the increase of yield strength by the Orowan mechanism can be expected to be $\sigma_0 \propto \sqrt{f}/d$ even if dispersoids are not homogeneously distributed but not very far away from this state. This is confirmed by the results shown in Fig. 4, where the yield strength increases linearly with the volume fraction of the increasing addition of SiC for both the room temperature and 300 °C.

Moreover, Fig. 4 displays an increase in yield strength of about 50% by an amount of 6 vol.% SiC addition for both temperatures, although the porosity of 1.5% is expected (see Fig. 4) and a noticeable amount of microcracks is present in the material.

Annealing experiments (see Fig. 5 and Fig. 6) show that the additional consolidation occurs at higher temperatures above 400 °C, but microcracks could not be completely removed, being confirmed by the damping measurements.

5. Conclusions

The yield strength of powder metallurgically produced aluminum can significantly be increased by Co-ball milling of aluminum powder with nanoscale SiC particles. The yield strength increase depends linearly on the volume fraction of added SiC particles as predicted by the Orowan hardening theory. After consolidation by hot extrusion the material contains microcracks that could not be totally removed by the heat treatment. Contrary to most aluminum alloys the produced aluminum matrix composites are sensitive for thermally produced cracks.

Acknowledgments

Financial support by Deutsche Forschungsgemeinschaft is gratefully acknowledged. The authors want to express their thanks to the referee of this article for his valuable and fruitful comments and clues.

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