INTERNAL FRICTION IN A Mg MATRIX COMPOSITE AFTER THERMAL CYCLING BETWEEN ROOM TEMPERATURE AND VARIOUS UPPER TEMPERATURES

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This paper describes the influence of thermal cycling on the damping behaviour of a Mg – 26.0 vol.% Al_2O_3 composite. The composite was prepared by squeeze casting. The logarithmic decrement as the characteristics of the process responsible for the damping behaviour was measured after ten thermal cycles between room temperature and an upper temperature. The upper temperature was varied in the interval from 100 to 450 °C.

The strain dependence of the logarithmic decrement can be divided into two regions. A region of strain amplitude independent damping at low strains is followed by a region with amplitude dependent damping at higher strains. The damping level in the strain amplitude independent region depends weakly on the upper temperature of the cycle whereas the values of the logarithmic decrement in the strain amplitude dependent region depend very sensitively on the upper temperature of the cycles.

The observed experiment data can be explained considering creation of new dislocations during thermal cycling.

VNITŘNÍ TLUMENÍ V KOMPOZITU NA BÁZI Mg PO TEPLOTNÍM CYKLOVÁNÍ MEZI POKOJOVOU TEPLOTOU A RŮZNÝMI VYŠŠÍMI TEPLOTAMI

V práci je studován vliv teplotního cyklování na vnitřní tlumení v kompozitu Mg – 26.0 obj.% Al_2O_3. Kompozit byl připraven metodou tlakového lití. Vnitřní tlumení je charakterizováno logaritmickým dekrementem útlumu. Dekrement byl měřen při pokojové teplotě po teplotním cyklování mezi pokojovou teplotou a horní teplotou cyklu. Horní teplota cyklu byla postupně zvyšována od 100 do 450 °C.

Amplitudovou závislost vnitřního tlumení je možné rozdělit do dvou částí. Pro malé amplitudy deformace je dekrement nezávislý, nebo slabě závislý na amplitudě. Pro am-

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plitudy vyšší než kritická amplituda je dekrement silně závislý na amplitudě. V oblasti amplitudově závislého dekrementu jsme také nalezli velký vliv teplotního cyklování. Pozorovaná experimentální data je možné vysvětlit vznikem nových dislokací při teplotním cyklování a přerozdělováním příměsí podél dislokačních čar.

1. Introduction

Dislocations and other lattice defects influence not only the mechanical behaviour of materials but also the damping response of the materials. The logarithmic decrement δ is often used to characterize the processes which determine the damping properties. The strain dependence of the logarithmic decrement may help to identify the physical processes responsible for the damping behaviour of the material. Magnesium is one of the materials where high damping is a result of the interaction between solute atoms and dislocations. Trojanová et al. [1] have reported some results on the investigation of the influence of Ca additions on the internal friction in Mg-Ca alloys prepared by rapid solidification and powder metallurgy. It has been shown that the internal friction measurements can be used as a non-destructive investigation of processes which influence the mechanical properties.

The search for metallic materials with higher specific strength led to the development of metal matrix composites (MMCs) which are candidates for various applications. Recently, Al, Mg, and Ti and their alloys as matrix reinforced with ceramic materials such as Al_2O_3, SiC, B_4C (in the form of discontinuous short fibres and particles) have been extensively investigated. It is known that in many cases there is a large difference in the coefficients of thermal expansion (CTE) between matrix and the reinforcement. When the MMC is cooled from a higher temperature to room temperature, misfit strains occur because of differential thermal contraction at the interface. These strains induce thermal (residual) stresses that may be higher than the yield stress of the matrix. Thermal stresses may be sufficient to generate new dislocations at the interfaces between the matrix and the reinforcements. Therefore, after cooling the composite, the dislocation density in the matrix increases, as proposed by Arsenault and Fisher [2] and observed using etch-pitting technique by Chawla and Metzger [3]. The dislocation generation due to relaxation of the thermal residual stresses may occur also during thermal cycling of MMCs. An increase in the dislocation density in the matrix after thermal cycling of Al-SiC composites has been demonstrated in an in situ transmission electron microscopy investigation by Arsenault and Shi [4]. Vogelsang et al. [5] investigated dislocation generation in 6061 Al alloy MMC reinforced with 20 vol.% SiC whiskers under thermal cycling using in situ transmission electron microscopy (TEM) observations. Very recently, the generation of dislocations in SiC particle reinforced magnesium alloy QE22 during thermal cycling has been investigated by Lukáč et al. [6] using acoustic emission (AE) technique. AE enables to detect the dislocation generation during thermal cycling in situ and it has advantage that AE is a non-destructive method. The high density of dislocations generated by the thermal stresses has been detected also during thermal cycling of Mg composites by Kiehn et al. [7]. These authors have also reported changes in the shape of specimens as a result of thermal cycling of MMCs.

One may expect that the dislocation generation and possible changes in the dislocation arrangement due to the difference in CTEs after thermal cycling may influence the damping behaviour of MMCs. The aim of this investigation is to determine the influence of the upper temperature of thermal cycles on the damping behaviour of a Mg-Al_2O_3 composite.

2. Experimental procedure

Commercial pure magnesium was used as matrix material. This was reinforced with δ-Al_2O_3 short fibres (Saffil) with a mean diameter of 3 μm and length about 87 μm. Composites were prepared by the squeeze casting method [8, 9]. The preforms consisting Al_2O_3 short fibres and a binder system (containing Al_2O_3 and starch) were pre-heated to a temperature higher than the melt temperature of magnesium (to about 1000 °C) and then inserted into a pre-heated die (290 to 360 °C). The pressure for forcing the melt into the die with the preform was applied in two steps (50 MPa for 10 s and 130 MPa for 60 s). The second step closes pores and shrinkage cavities. During this short time of a contact between the liquid metal and the fibres, only a slight reaction between the fibres and the matrix can occur. Mg_2Si and $Mg_{17}Al_{12}$ are formed at the matrix-fibre interfaces due to reactions of the melt with the Saffil fibres and/or the binder. Diffractometer phase analysis gave some evidence for the presence of MgO. The two-stage application of the pressure resulted in MMCs with a fibre volume fraction of 26.0 vol.% fibres.

Test specimens for the damping measurements were machined as bending beams (88 mm long with thickness of 3 mm) with the reinforcement plane perpendicular to the main specimen axis. The damping measurements were carried out in vacuum (about 30 mPa) at room temperature. The specimens fixed at one end were excited into resonance (the frequency ranged from 200 to 250 Hz) by a permanent magnet and a sinusoidal alternating current in the exciting coil. Damping was measured as the logarithmic decrement δ of the vibrating beam. The signal amplitude is proportional to the strain amplitude ε. A special algorithm using all points was used for calculation of the strain amplitude dependence of the logarithmic decrement. Experimental details are described by Buchhagen et al. [10]. Thermal cycling was carried out using a radiant energy furnace. Forced gas (air) was supplied during the cooling phase of thermal cycling. Ten thermal cycles between room temperature and an upper temperature were performed. The upper temperature was varied in the temperature range from 100 to 450 °C.

3. Results and discussion

Damping of materials depends generally on stress and strain. Fig. 1 shows the plots of the logarithmic decrement against the logarithm of the strain amplitude for Mg-Saffil composites before and after ten thermal cycles between room temperature and various higher temperatures. It can be seen that the strain dependencies of the logarithmic decrement exhibit two regions. In the first region, for lower strain amplitudes, the logarithmic decrement is practically independent of the maximum strain amplitude. In the second region, for higher strains, the logarithmic decrement depends on the strain amplitude. A very strong strain dependence of the logarithmic decrement is measured for MMCs after thermal cycling. The value of the amplitude dependent logarithmic decrement for MMCs before thermal cycling is much lower than that for MMCs after thermal cycling. The values of δ in the strain amplitude dependent region increase very strongly with increasing upper temperature up to 300 °C and then, above 300 °C, the values of δ decrease with the upper temperature. The value of the critical strain ε_c at which the logarithmic decrement begins to increase with the strain amplitude for MMCs after thermal cycling is much lower than that for MMCs before thermal cycling. The critical strain decreases with increasing upper temperatures below 300 °C and then it is practically independent of temperatures between 300 and 450 °C. It should be noted that no pronounced dependence of the damping behaviour on the number of cycles can be detected up to 100 cycles [7]. Trojanová et al. [11] have reported qualitatively similar strain amplitude dependencies of the logarithmic decrement for magnesium reinforced by 20 vol.% Saffil prepared by squeeze casting but after a heat treatment.

The strain amplitude dependence of the logarithmic decrement suggests dislocations unpinning processes. The differences in the damping behaviour between specimens thermally cycled to various temperatures can be attributed to the interaction between dislocations and point defects including small clusters of foreign atoms and to changes in the dislocation density.

The observed behaviour may be explained if we consider that during cooling and also during thermal cycling new dislocations are created due to the difference in the CTE and/or that new pinning points on existing dislocations are formed owing to reactions between the matrix and the reinforcement.

The strong strain dependence of the logarithmic decrement for Mg-Saffil composite specimens shown in Fig. 1 may be explained using the Granato-Lücke theory of dislocation damping [12]. According to this theory the dislocation segments are pinned by weak and strong pinning points. The mean distance between weak and strong pinning points is l and L, respectively. At low applied strain amplitudes the dislocation segments bow out between the weak pinning points which are produced by impurity atoms and/or small precipitates and remain anchored at the strong pinning points. With increasing strain amplitude the force on the dislocation seg-

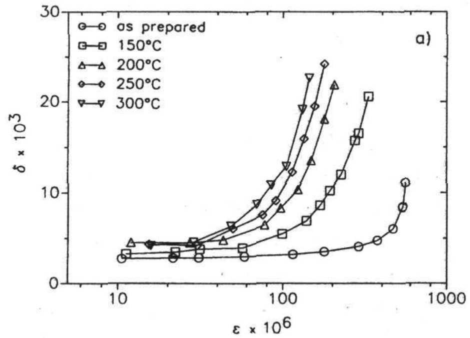

Fig. 1a. Strain amplitude dependence of the logarithmic decrement for Mg – 26 vol.%. Saffil cycled between room temperature and an upper temperature up to 300 °C.

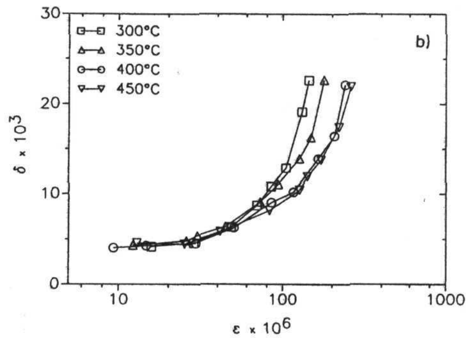

Fig. 1b. Strain amplitude dependence of the logarithmic decrement for Mg – 26 vol.%. Saffil cycled between room temperature and an upper temperature from the range of 300 to 450 °C.

ment becomes higher than the binding force F_B between weak pinning points and a dislocation. The dislocation segments break away from some weak pinning points. This leads to an instantaneous increase in dislocation stress and thus gives rise to an increased level of the damping. The logarithmic decrement can be expressed as a sum of two components

$$\delta = \delta_0 + \delta_H, \tag{1}$$

where δ_0 represents the strain independent component of the logarithmic decrement and δ_H is the strain dependent component of δ. The both components can be expressed by [12, 13]

$$\delta_0 = C_0 \rho L_e^4 , \tag{2}$$

where C_0 is a constant, ρ is the density of dislocations and

$$1/L_e = 1/L + 1/l \tag{3}$$

and

$$\delta_H = (C_1/\varepsilon) \exp(-C_2 \varepsilon) \tag{4}$$

with

$$C_1 = F_B \rho L^3 / 6bEl^2 \tag{5}$$

and

$$C_2 = F_B / bEl , \tag{6}$$

where b is the magnitude of the Burgers vector of the dislocations and E is the unrelaxed modulus. From Fig. 2 it can be seen that the values of C_2 depend on the upper temperature of the cycles.

The observed increase in the logarithmic decrement in Mg-Saffil MMCs after thermal cycling in comparison to these composites before thermal cycling (at the same strain amplitude) may be explained if we consider that the density of dislocations after thermal cycling is higher than the dislocation density in as-prepared specimens. Likewise l and/or L may increase due to thermal cycling. It is well known that there is a difference in the CTE of the matrix and that of the reinforcement. The difference in the CTE values of both composite components falls into a range 10–20 × 10^{-6}K^{-1}. It is expected that such a large difference in the CTE values generates thermal stresses at the interfaces between the Mg matrix and alumina fibres during thermal cycling. The thermal stresses induced by thermal cycling may be of sufficiently high magnitude to generate dislocations around reinforcements. Generation of new dislocations owing to the thermal expansion mismatch between the matrix and the reinforcement has been reported in specimens under thermal cycling [5–7].

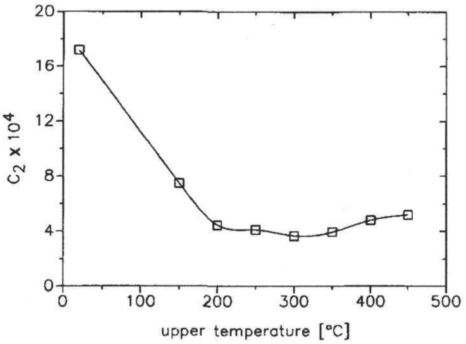

Fig. 2. Temperature dependence of the constant C_2.

The dislocation density produced by the thermal stress can be calculated as [4, 14]

$$\rho = \frac{B \cdot \Delta\alpha \cdot \Delta T \cdot V_f}{(1 - V_f) \cdot b \cdot t}, \tag{7}$$

where B is a geometric constant, ΔT is the temperature difference, $\Delta\alpha$ is the difference in the CTEs, V_f is the volume fraction of the reinforcement, and t is the smallest dimension of the reinforcement. The flow stress corresponding to the dislocation density ρ can be estimated as

$$\sigma = AGb\rho^{1/2}, \tag{8}$$

where A is a constant and G is the shear modulus. If the thermal stresses reach a certain critical value, dislocations can move. The dislocation motion causes plastic deformation resulting in residual elongation, which was observed experimentally [7, 15].

We consider that thermal stresses contribute to the stresses induced by the internal friction apparatus. The force acting on the dislocation segments becomes higher than the binding force and the dislocations can breakaway from the pinning points. Above a certain temperature the thermal stresses are higher than the yield stress of the matrix and hence, the dislocation generation and motion can start during thermal cycling. According to Eq. 7, the density of the new created dislocations, due to the difference between the CTEs of the matrix and the fibre, increases with increasing upper temperature of the thermal cycle. The higher the upper temperature the more dislocations are generated. The total dislocation density increases. On the other hand, the concentration of the weak pinning points (corresponding to the concentration of impurity atoms) remains constant. It means that l increases effectively with increasing upper temperature. The increase in the logarithmic decrement with increasing temperature at the same strain amplitude may be explained if we take into account that l and ρ increase with increasing upper temperature. An increase in l results in a decrease in the value of C_2 according to Eq. 6, which is observed. As the distance between weak pinning points increases effectively with the upper temperature, the dislocations can breakaway from the weak pinning points under lower stresses and thus the critical strain should decrease with increasing upper temperature which is observed (see Fig. 1). An increase in the dislocation density ρ as well as in l with increasing upper temperature influences the value of C_1. The value of C_1 should increase with increasing ρ but it should decrease with increasing l. Both ρ and l increase with increasing upper temperature and hence, a monotonous variation of C_1 with the upper temperature cannot be expected, which is observed.

The thermal residual stresses may relax due to the movement of dislocations. The development of a refined substructure and rearrangement of dislocations under the thermal forces during thermal cycling may be expected. Therefore, thermal cycling has, in comparison to the cooling, a more significant influence on the microstructural changes in the matrix due to creating and moving dislocations. These processes may also result in an increase of L. It is very probable that a stress-assisted thermally activated redistribution of the weak pinning points may occur during thermal cycling. The distance between the weak pinning points l should increase. The dislocations can break away. The changes in the dislocation density, L, and l due to thermal cycling influence both the value of the logarithmic decrement which is higher and the critical strain which is lower for specimens after thermal cycling as compared to specimens before thermal cycling. As the critical strain corresponds to the microyield stress, an increase in l causes a decrease in the critical strain because the stress required for the dislocation breakaway from the weak pinning points is inversely proportional to l. However, the stress necessary for the dislocation breakaway decreases with increasing temperature and hence, thermal cycling may make easy the breakaway of dislocations. On the other hand, at higher temperatures annihilation of dislocations can occur during thermal cycling which implies an increase of L and therefore, an increase in the logarithmic decrement in composites after thermal cycling.

It should be noted that an analysis using finite element models [16] shows that the residual strain which develops during cooling depends not only on the temperature difference but also on fibre spacing. According to this analysis, the fibre distribution at a certain fibre volume fraction does not have a significant effect on the thermal stresses. It is obvious that the dislocation movement is determined by the required stress that depends on temperature, the distribution and kind of obstacles (pinning points), the internal stresses and the crystallographic orientation of the grains to the fibres. The dislocation density near the interface between the fibres and the matrix is expected to be significantly higher than elsewhere in the matrix. Thus a plastic zone around the fibres containing tangled dislocations may be formed. The interface between the fibre and the matrix is one of the important characteristics of the microstructure. Further detailed investigations are required to understand better the role of the fibre-matrix interface and dislocations in the damping behaviour of MMCs.

4. Conclusion

The values of the logarithmic decrement, as the characteristic quantity of damping, for the magnesium matrix composites after thermal cycling are much higher, especially in the strain amplitude region, than those for the composite specimens before thermal cycling. The values of the logarithmic decrement depend on the upper temperature of a thermal cycle.

Magnesium matrix composites reinforced by Saffil fibres, as candidates for structural materials because of their better specific strength than conventional alloys, exhibit a relatively high value of damping. They may be considered as high damping materials, even if the damping values are lower than those for cast pure magnesium [17, 18]. In engineering applications the ability of a material to dampen out mechanical vibrations is a very useful property. Hence, it is important to investigate their damping behaviour which can be influenced by thermal treatment.

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