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Effects of macroscopic graphite particulates on the damping behavior of CuAlMn shape memory alloy

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Abstract A novel composite was fabricated using a Cu–11.9Al–2.5Mn (wt%) shape memory alloy as the matrix and macroscopic graphite particulates as the second phase. The damping behavior of the resultant composite was examined in the present study through internal friction measurements. It was found that the net height of the internal friction peak relating to phase transformation decreased in the composite due to the constraining effect of the particulates, but the internal friction background significantly increased with increasing the volume fraction or decreasing the diameter of the particulates, giving rise to an improved damping capacity particularly at low temperatures. It is proposed that the interface damping, dislocation damping and the intrinsic damping of the matrix and the constraint are predominant in the composite.

Introduction

High damping capacity has been one of the most important properties of materials used in engineering structures where undesirable noise and vibration are to be passively attenuated. Among the prevalent high damping metallic materials, shape memory alloys (SMA) could be one of the most promising candidates due to their very high damping capacity arising from the reversible martensitic transformation (MT) and the stress induced reorientation of martensite variants [1, 2].

It is generally accepted that micro-structural defects should play a dominant role in the damping response of materials [3–5]. From this fundamental concept, a variety of high damping metals and metal matrix composites (MMCs) have been developed. Reinforcements, however, have been shown to provide limited improvements in the resultant damping capacity, since the quantities of reinforcement are limited. It has been disclosed in our previous studies that the secondary stable phase can be either particulates or pores in macroscopic dimension if damping enhancement is required [6, 7]. The damping capacity of alloys, with either a high or low intrinsic damping, can be substantially improved through this macroscopic MMC route, typically from several times to an order higher than that of the parent materials. It is therefore rationalized that the damping behavior of SMA can also be tailored in this way to get increased damping together with decreased density. Such novel MMC can find use in a variety of industrial areas, including lightweight structures and energy absorption apparatus, etc. However, the functional properties of such novel MMC should be well understood before they can be considered for industrial applications because the SMA serving as the matrix of the composite must be very different from usual metals. The functional properties are directly related to the constraining behavior that the macroscopic reinforcement has on the SMA matrix, and that the constrained thermo-elastic martensitic transformation is very different from that in a free condition [8–10]. The investigation of this topic is of particular

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interest because the thermo-elastic martenstic transformation is the basis of the shape memory effect (SME) and pseudoelasticity (PE) of SMAs. Recently, researchers have looked at the martensitic transformation behavior of SMAs under a constraint condition based on differential scanning calorimetry (DSC) or mechanical property experiments [8–13]. Nevertheless, it is interesting to note that in the literature there has shown very little respect to the damping behavior associated with the martenstic transformation or the modification of the microstructures of the SMA under a constrained condition. And the tool of internal friction has not been adopted, either. In the present study a novel MMC has been fabricated with a high volume fraction of macroscopic reinforcement aimed at giving a clear physical image of the damping behavior and correlated mechanism of the novel MMC with the objective of understanding the transformational behavior of SMA under a constrained condition.

As a part of a series of investigations, the present work is focused on the damping behavior of a Cu-based SMA reinforced by macroscopic graphite particulates. It is expected that the present study may give guidelines to improve materials' damping capacity and to understand correlated mechanisms.

Experimental procedure

Preparation of the metal matrix composite (MMC) Specimens

The air pressure infiltration process was employed to fabricate the composite specimens with a Cu–11.9Al–2.5Mn (wt.%) SMA and graphite particulates (Gr). The graphite particulates with a chosen size were weighed using an electronic balance with an accuracy of 0.01 g and uniaxially pressed into a steel die with an inner diameter of 70 mm under an appropriate pressure. The porous compact was then preheated to 1173 K and infiltrated with the SMA melt under one atmosphere pressure, resulting the required MMC ingots. Both the composite and the unreinforced bulk specimens used for comparison were cut from the same ingot using an electric sparking machine to guarantee the identity of microstructures among all the specimens. Before the damping measurements, all the specimens were quenched with water after holding at 1173 K for 900 s.

The Specimen density was determined from its weight and physical dimensions, from which the volume fraction of the graphite particulates was calculated by the following equation

$$V_G = \frac{\rho_A - \rho_C}{\rho_A - \rho_G}, \tag{1}$$

with V_G as the volume fraction of macroscopic graphite particulates; ρ_A, ρ_C and ρ_G as the densities of the matrix, the composite and the bulk graphite, respectively.

Damping measurement

Internal friction (IF), Q^{-1}, was used to characterize the damping property of the specimens and was measured using a computer controlled automatic inverted torsion pendulum by forced vibration. This apparatus basically consists of an inverted torsion pendulum, a temperature programmer and a photoelectron transformer. The whole measurement is controlled by an IBM*PC586 computer and an 8087 processor and the data can be processed in real time. The range of the maximum excitation strain amplitude is 10^{-6}–10^{-4}. The resolution in the IF measurements is 1×10^{-4}. The details of the experimental apparatus can be found in the references [6, 7].

The pore distribution and pore size were analyzed with optical microscopy. The microstructures of the specimens were characterized with TEM and Field Emission SEM.

Results and analysis

Damping behavior of the bulk graphite and CuAlMn alloy

In order to aid in analyzing the damping mechanisms of the composite, the inherent damping property of each constituent was first examined. Figure 1 exhibits the IF as a function of temperature for the bulk graphite. It is seen that the IF is essentially temperature independent and keeps a mean value of around 0.016 up to 623 K, being in good agreement with reported values [14]. The damping mechanisms of bulk graphite have been demonstrated to be mainly attributed to sweeping motion of high density glissile dislocations and sliding between the basal planes under cyclic loading [15].

Figure 2 exhibits the IF and relative dynamic modulus (RDM) as the functions of temperature for the bulk CuAlMn alloy. The most typical feature would be the presence of two IF peaks at about 423 K (termed P1) and 598 K (termed P2), respectively, with similar dependence on vibration frequency. The RDM roughly

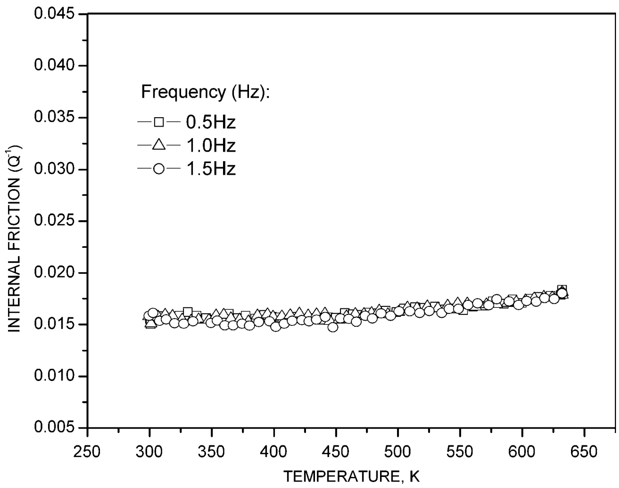

Fig. 1 The essential IF-temperature spectra of the bulk graphite

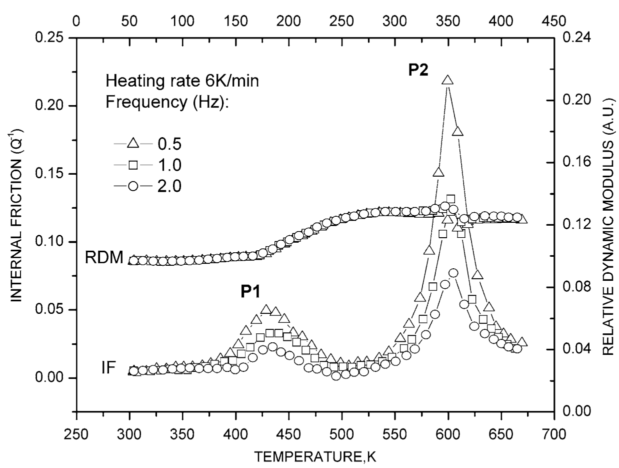

Fig. 2 The essential IF-temperature spectra of the bulk CuAlMn SMA

keeps unchanged from room temperature up to P1 temperature and thereafter it starts to rapidly increase until the occurrence of the P2 peak, where a discontinuity takes place. The presence of the P1 peak is confirmed to originate from the refinement of the twins during heating, belonging to a pseudo first order transition [16, 17].

As shown in Figs. 2 and 3, the P2 peak rises as measuring frequency decreases or heating rate increases. The location of the peak slightly shifts towards high temperature with increasing heating rate while keeps unchanged with changing measuring frequency. Moreover, both the peaks show a typical feature of hysteresis during cooling as shown in Fig. 4. All these features accord well with the general characteristics of the first-order phase transition. Combination of these

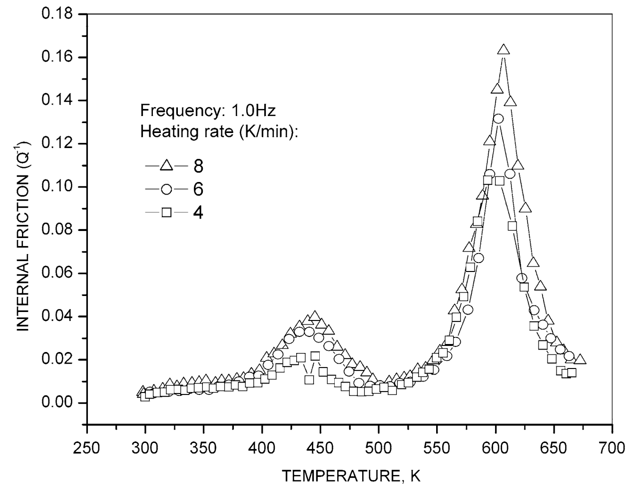

Fig. 3 Influence of heating rate over the P1 and P2 peaks of the bulk CuAlMn SMA

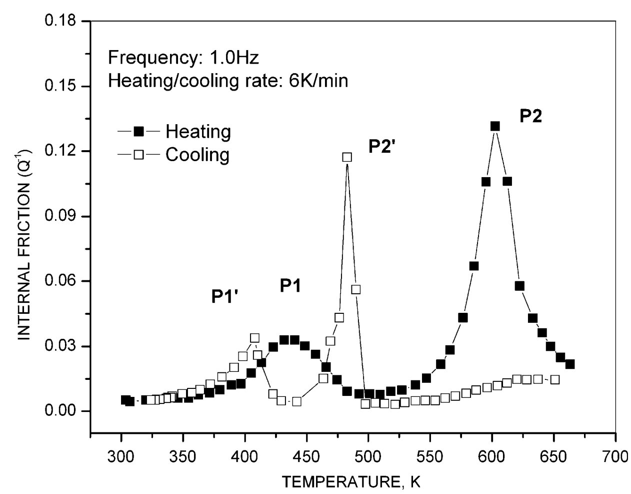

Fig. 4 Hysteresis phenomenon of the IF peaks during heating and cooling

features with the results of differential scanning calorimetry (DSC) shown in Fig. 5, it is suggested that the P2 peak arises from the reverse MT.

Effect of the graphite particulates on the damping behavior of the CuAlMn SMA

Figure 6 shows the distribution of the macroscopic graphite particulates in the composite as well as the microstructures of the matrix in the bulk alloy and composite. The graphite particulates have a granular shape uniformly distributed over the matrix and the composite matrix exhibits a relatively thin martensitic structure consisting of twin variants compared with the bulk alloy.

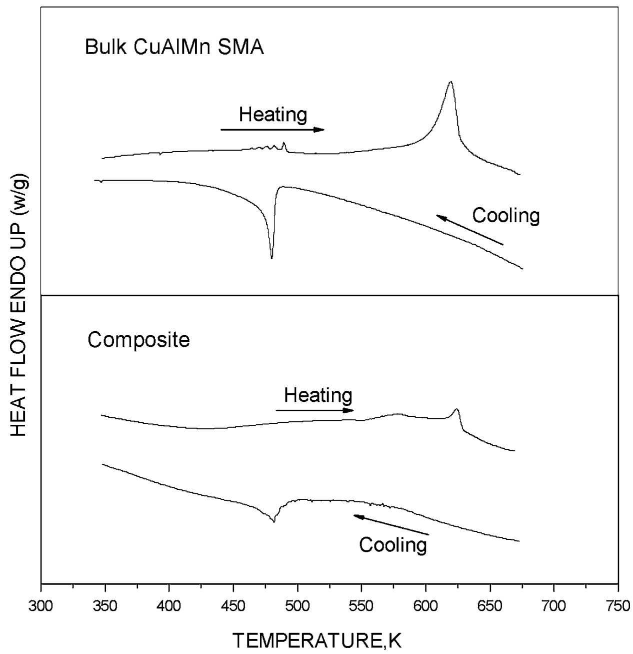

Fig. 5 DSC results for the composite and bulk CuAlMn SMA measured at a heating/cooling rate of 10 K/min

Figure 7 illustrates the comparison of the IF against temperature between the bulk alloy and the composite reinforced with varied volume fractions of the graphite particulates. The most significant changes in the composites should be the elevation of IF background, i.e. the overall damping capacity of the composite is enhanced by the graphite particulates. For instance, compared with the bulk SMA that has an average IF of 0.006 at ambient temperature, the composite with 74% reinforcement shows an average IF value of 0.028, 4 times higher than that of the bulk alloy. The enhancement becomes more pronounced as the volume fraction of the particulates increases or the particulate size decreases, as shown in Fig. 8. However, it should be noted that all the composites show decreased net heights, i.e. after the background IF subtracted, of both P1 and P2 peak.

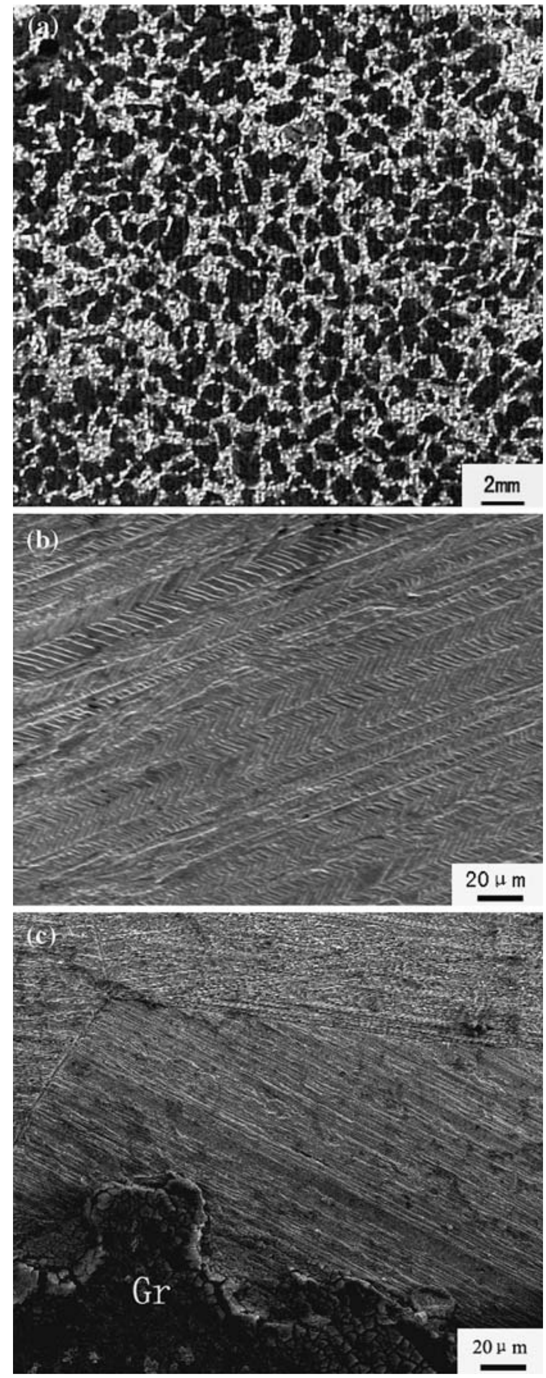

Fig. 6 Characterization of the microstructures of the SMA with (**a**) distribution of graphite particulates over the SMA matrix and (**b**) martensite structure of the bulk alloy and (**c**) martensite structure of the composite

Discussion

Contribution of each constituent to the overall damping of the composite

For particulates reinforced MMC, the operative damping mechanisms have been extensively investigated and can be described by the extended rule of mixture (ROM) [18], that is

$$\psi_c = \psi_m V_m + \psi_p V_p + \psi_i V_i, \tag{2}$$

where ψ_c, ψ_m and ψ_p and ψ_i represent the damping capacity of the composite, matrix and particulates and particulate/matrix interfaces, respectively; V_m, V_p and V_i denote the volume fraction of the matrix, particulates and particulate/matrix interface, respectively, where the particulate/matrix interface is actually

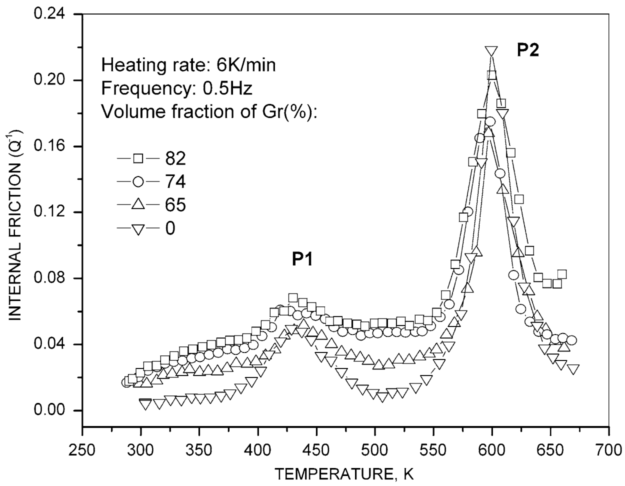

Fig. 7 Effect of the volume fraction of the graphite particulates on the IF of the CuAlMn SMA

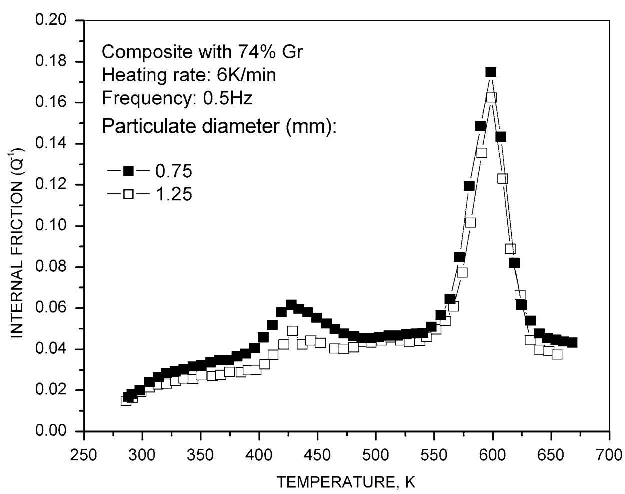

Fig. 8 Effect of the graphite particulate size on the IF of the CuAlMn SMA

composed of distorted atom layers or domains with a certain thickness in both the particulates and the matrix and thus is recognized as a ‘‘volume’’. From this relationship, it is concluded that the overall damping of a composite is not only correlated to the individual damping capacity of each constituent but also to the interface.

Figure 9 gives a comparison between the measured IF data of the composites and the theoretical ones obtained by calculating only the former two terms of the Eq. (2), i.e. the contributions of the matrix and the reinforcement. The difference between the two curves that reflects the contribution of the interface to the overall IF of the composites increases as the volume fraction of the reinforcement is raised. Moreover, the

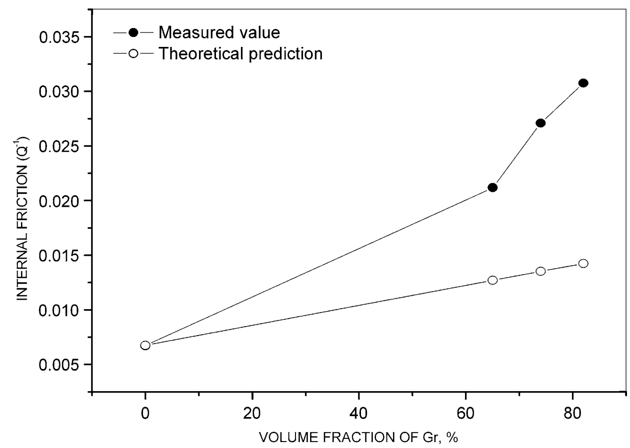

Fig. 9 Comparison between the measured IF values (filled cycles) for the composite and the theoretical predictions (open cycles) calculated by extended ROM for the matrix and the graphite particulates

difference is larger than the IF sum of the matrix and the reinforcement at each particulate content, suggesting that the interface does play a predominant role in the overall damping among the three components. Of course, some changes have actually taken place in the matrix microstructure of the composite due to the existence of graphite particulates. These embedded graphite particulates create a constrained circumstance for the nucleation and growth of the martensitic variants, leading to relatively fine microstructures, as compared in Fig. 6(b) and (c). The fine martensitic variants are obviously favorable for the elevation of IF background in terms of the damping mechanisms of SMA.

Interface damping

In addition to the modification of the matrix microstructure, considerable particulate/matrix interfaces and thermal mismatch induced dislocations in the adjacent matrix arise in the composite [4, 19–23]. It has been confirmed that the macroscopic particulate reinforced metal matrix composites prepared by the infiltration process basically possesses a weakly bonded reinforcement-matrix interface [8, 24] and on the other hand, the interfacial bonding is likely to be weak due to the layered faces structure presenting in graphite particulates.

The effect of weakly bonded interfaces on the overall damping capacity of particulates reinforced composites has been systematically investigated by a number of researchers using interface slip model [25, 26]. In this model, interface damping is proposed to arise from the frictional energy loss between the particulates and the matrix when the magnitude of applied shear stress is sufficient to overcome the frictional resistance. In the case of weakly bonded interfaces, the interfacial slip is relatively easy to occur due to relatively weak frictional resistance. The resultant damping is therefore higher and is a major damping source in composites.

Dislocation damping

In particulate reinforced composites, the particulate/matrix interface is a two-dimensional defect where the crystal structure of the matrix is distorted, leading to a complicated stress–strain state in the area near the interface. This state will be even more serious for the present composite because the SMA matrix undergoes a reversible MT during heating and cooling. When the composite solidified from its processing temperature, stresses will be generated due to the difference in the thermal expansion coefficients (CTEs) of the matrix and reinforcement [23, 27]. The stress can plastically deform the metal matrix surrounding the reinforcement, leaving residual thermal mismatch strain, ε, which is the function of the difference between the CTEs of the reinforcement and matrix, $\Delta\alpha$, that is

$$\varepsilon = \Delta\alpha\Delta T, \tag{3}$$

where ΔT is the variation of temperature.

The CTE value of the present CuAlMn alloy is approximately $16 \times 10^{-6}\ K^{-1}$ and that of the graphite is approximately $4.0 \times 10^{-6}\ K^{-1}$ [28, 29]. The processing temperature of the present composite specimens is 1373 K and thus the temperature difference ΔT between it and room temperature can be assumed to be 1075 K. By substituting the values of ΔT and CTE into Eq. (3), it is found that the thermal mismatch strain at the matrix/reinforcement interface is around 1.29%. The estimated stress in the neighboring matrix at this strain is approximately 900 MPa, much higher than the yield strength of the CuAlMn alloy, i.e. 300 Mpa [30]. As a consequence, microplastic deformation occurs although those thermal mismatch stresses will be partially relaxed during the phase transformation, making the dislocations move and new dislocations form. This increased dislocation density has been observed adjacent to the particulate/matrix interfaces, as shown in Fig. 10. Furthermore, it is well known that if the damping originates from a dislocation mechanism, it should be strain amplitude dependent. The results in Fig. 11 do show the dependence, i.e. the IF increases as the strain amplitude increases, demonstrating the existence of dislocation damping.

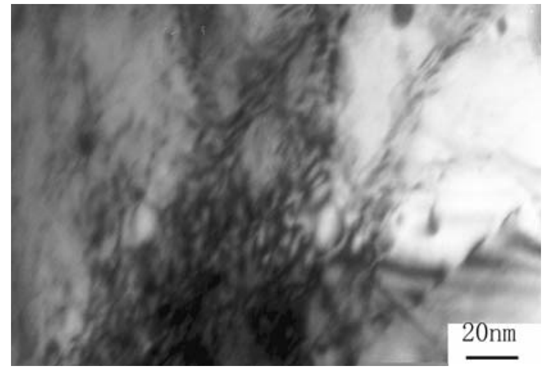

Fig. 10 Configuration of dislocations in the matrix adjacent to the graphite particulates

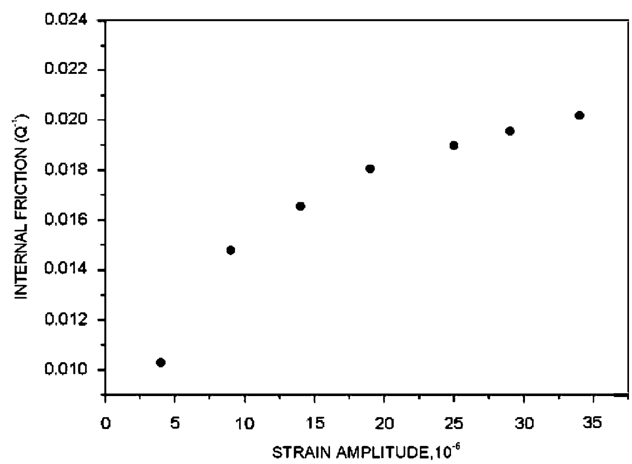

Fig. 11 The effect of strain amplitude on the IF of the composite

Influence of the graphite particulates on the IF peaks

From aforementioned results, although the IF background of the composite was raised compared with that of the bulk alloy, the net height of the P2 peak arising from the phase transformation decreased in the composite. This phenomenon can be tentatively ascribed to the occurrence of so called recovery stress during reverse MT due to the impediment of the constraint that the macroscopic graphite particulates have on the SMA matrix [9, 11, 31, 32]. It has been known that the thermo-elastic martensitic transformation is sensitive to both stress and temperature and thus the constrained phase transformation of SMAs should therefore be very different from that in a free condition. According to published reports [8, 9, 11], the differences are mainly due to the coexistence of two kinds of martensites, i.e. self-accommodating martensite (SAM) and preferentially oriented martensite (POM), in the constrained condition. SAM-variants are formed if the forward transformation occurs under a free condition. The transformation strain together with the formation of the variant can be compensated by the surrounding variants so that the global net strain is zero. If an external stress is applied to SAM-variants, they will transform into POM-variants and a net strain is produced. When heated in a free condition, the pre-strained SMA undergoes a reverse MT from SAM and POM variants into the parent phase and makes the shape memory strain recovered. If some constraints are employed during heating, however, the transformation of SAM will not be impeded because no macroscopic shape change takes place while that of the POM will be opposed due to the occurrence of a net strain during the transformation that acts on the constraint. As a result, recovery stresses are generated, leading to an incomplete transformation from the POM to the parent phase even though heated to a much higher temperature than A_f, i.e. the finishing temperature of reverse transformation. As aforementioned that when the composite solidified from its processing temperature the thermal mismatch strain at the matrix/reinforcement interface is around 1.29% and the estimated stress in the neighboring matrix at this strain is approximately 900 Mpa. As a result, some POM variants may be formed near the matrix/reinforcement interfaces. When heated to the reverse MT temperature, the transformation of the POM-variants into the parent phase is impeded because of the recovery stress, leading to decreased P2 peak. To clarify the mechanisms of the impediment of macroscopic reinforcement on the MT more clearly, further research is in process.

From Fig. 7, on the other hand, the net height of the P1 peak of the composite is also lower than that of the bulk alloy. This means that the constraint resulted from the graphite particulates restrains not only the MT but also the sliding of the twins through producing dislocations as well as other defects like voids. The decrease of the P1 peak would be understandable.

Conclusions

The damping behavior of a novel composite fabricated with macroscopic graphite particulates and a CuAlMn shape memory alloy has been investigated in the present study. Two peaks were found on the internal friction-temperature spectrums of the composite and the bulk alloy, in which the high-temperature peak was demonstrated to be related to the reverse martensitic transformation while the low-temperature one arises from the refining of twins. The composite shows an enhanced internal friction background but decreased internal friction peaks. The enhancement becomes more pronounced with increasing the volume fraction or decreasing the diameter of the particulates. It is proposed that the interface damping, dislocation damping and the inherent damping of the constituent phases in the composite are responsible for the damping enhancement.

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