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Cu-Al-Ni-SMA-Based High-Damping Composites

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Recently, absorption of vibration energy by mechanical damping has attracted much attention in several fields such as vibration reduction in aircraft and automotive industries, nanoscale vibration isolations in high-precision electronics, building protection in civil engineering, etc. Typically, the most used high-damping materials are based on polymers due to their viscoelastic behavior. However, polymeric materials usually show a low elastic modulus and are not stable at relatively low temperatures (≈323 K). Therefore, alternative materials for damping applications are needed. In particular, shape memory alloys (SMAs), which intrinsically present high-damping capacity thanks to the dissipative hysteretic movement of interfaces under external stresses, are very good candidates for high-damping applications. A completely new approach was applied to produce high-damping composites with relatively high stiffness. Cu-Al-Ni shape memory alloy powders were embedded with metallic matrices of pure In, a In-10wt.%Sn alloy and In-Sn eutectic alloy. The production methodology is described. The composite microstructures and damping properties were characterized. A good particle distribution of the Cu-Al-Ni particles in the matrices was observed. The composites exhibit very high damping capacities in relatively wide temperature ranges. The methodology introduced provides versatility to control the temperature of maximum damping by adjusting the shape memory alloy composition.

Keywords Cu-Al-Ni shape memory alloys, high damping, In-Sn, metal matrix composites

1. Introduction

In the last decade, absorption of vibration energy by mechanical damping has attracted much attention in several fields such as vibration reduction in aircraft and automotive industries, nanoscale vibration isolations in high-precision electronics, building protection in civil engineering, etc. (Ref 1-5). For the design of structural applications, materials combining both a high-damping capacity and high stiffness at moderate temperatures are required, but, unfortunately, to find materials with these characteristics is not frequent (Ref 6-9). Indeed, the merit index for damping materials is a combination of the elastic modulus (E) and the damping coefficient tan ϕ, where ϕ is the loss angle between the strain experienced by the sample and the stress applied (Ref 6, 7). Typically, polymers were used as high-damping materials due to their viscoelastic behavior (Ref 7, 8). However, polymeric materials usually show a low elastic modulus and are not stable at relatively low temperatures (≈323 K). These drawbacks have encouraged the scientific community to look for alternate materials for damping applications.

High damping metallic materials (HIDAMETS), which, in principle, exhibit higher modulus and thermal stability than polymers, with similar damping properties were proposed (Ref 9). Another approach was the development of metal matrix composites, in which each component or phase has a specific role: damping or stiffness (Ref 10). Among HIDAMETS, shape memory alloys (SMA) intrinsically show very high damping properties thanks to the dissipative hysteretic movement of interfaces under external stresses (Ref 11-13) and have found some practical applications (Ref 4, 5, 14-17).

Making use of both the high-damping capacity of Cu-Al-Ni SMAs and the flexibility of composite production, high-damping metal matrix composites were developed (Ref 18, 19). Cu-based SMAs were used because they definitively exhibit higher damping capacity and are not so expensive as the commercially widespread Ni-Ti ones. Phenomena like interdiffusion, intermetallic formation, or grain boundary penetration were observed in the different composites studied (Ref 19, 20).

The processing method to elaborate these materials as well as their microstructure and damping properties will be reviewed in the present work. Results on the damping behavior will be discussed as well.

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2. Experimental Procedures

Cu-Al-Ni SMA powders produced by Ar atomization (Ref 21) with two different compositions and a particle diameter of 25-50 μm were embedded with metallic matrices by infiltration (Ref 18, 19). Pure In, a In-10wt.%Sn alloy, and a eutectic In-Sn alloy were employed as matrices. To produce the composites, firstly, a weighted amount of powders of the corresponding Cu-Al-Ni SMA was placed at the bottom of a

Teflon mould under vacuum ($p \approx 13.3$ Pa). Pieces of the matrix material were also suspended inside the tube. The assembly was heated up to a temperature below 500 K, depending on the particular matrix, and subsequently a moderate pressure (0.3-0.4 MPa) for a short time (≈10 min) was applied. Finally, the assembly was cooled down in water. More details can be found elsewhere (Ref 18, 19). Relevant data for the SMA powders, matrices, and composites are summarized in Table 1. The transformation temperatures, i.e. martensite start (M_s) and martensite finish (M_f) and the analogous ones for austenite (A_s and A_f), of the as-atomized powders determined by differential scanning calorimetry are also given. The volume fraction of the embedded SMA powders was determined by image analysis.

Specimens were prepared for metallographic analysis by standard procedures. The final polishing step was performed using colloidal silica suspension. Optical microscopy (OM) was performed in a Leica DMRXA equipped with a Nomarski prism. Differential scanning calorimetry (DSC) measurements were performed using a Perkin–Elmer DSC7 equipment.

The damping behavior of the composites was investigated by mechanical spectroscopy using a forced inverted torsion pendulum (Ref 22, 23).

Table 1 Data of the SMA powders, matrices, and composites produced

	Composite		
	A	B	C
Powder composition, wt.%	Cu-13.1 Al-3.1Ni (P1)	Cu-14.15 Al-3.3Ni (P2)	Cu-14.15 Al-3.3Ni (P2)
Transformation temperature of the as-atomized powders, K			
M_s	338	253	253
M_f	300	206	206
A_s	324	220	220
A_f	368	289	289
Matrix	Pure In	In-10wt.%Sn	In-49.1wt.%Sn (eutectic)
SMA, vol.%	61 ± 1.5	57 ± 2	60 ± 2

3. Experimental Results and Discussion

3.1 Composite Microstructure

Images of the composites microstructures acquired using polarized light in the optical microscope at room temperature are presented in Fig. 1. In the overview of composite A (Fig. 1a), it is noticeable how the Cu-Al-Ni SMA particles are well surrounded by the In matrix. All the Cu-Al-Ni particles have undergone the martensitic transformation after the composite production and are in martensite state. This was expected, because the transformation temperatures of the as-atomized powders used in this composite are above room temperature (see Table 1). According to the image analysis applied to several images of this composite, there is a 61 vol.% of SMA particles.

A similar microstructure was observed for composite B (Fig. 1b), where Cu-Al-Ni SMA particles, with composition P2, were embedded by a In-10wt.%Sn matrix in a uniform way too. Moreover, irregularly shaped particles are very well impregnated with the matrix even in intricate zones. The polycrystalline character of the SMA particles in the austenitic state can be appreciated clearly. The amount of SMA particles was determined to be 57 ± 2 vol.%. The occurrence of

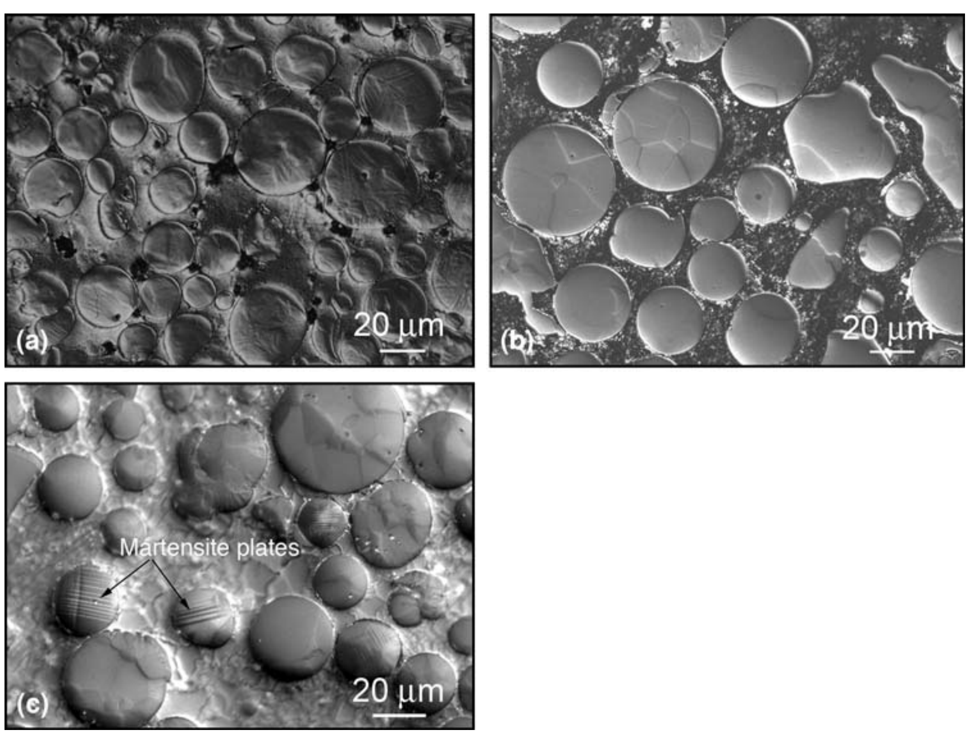

Fig. 1 Optical micrographs of composite A (a), B (b), and C (c)

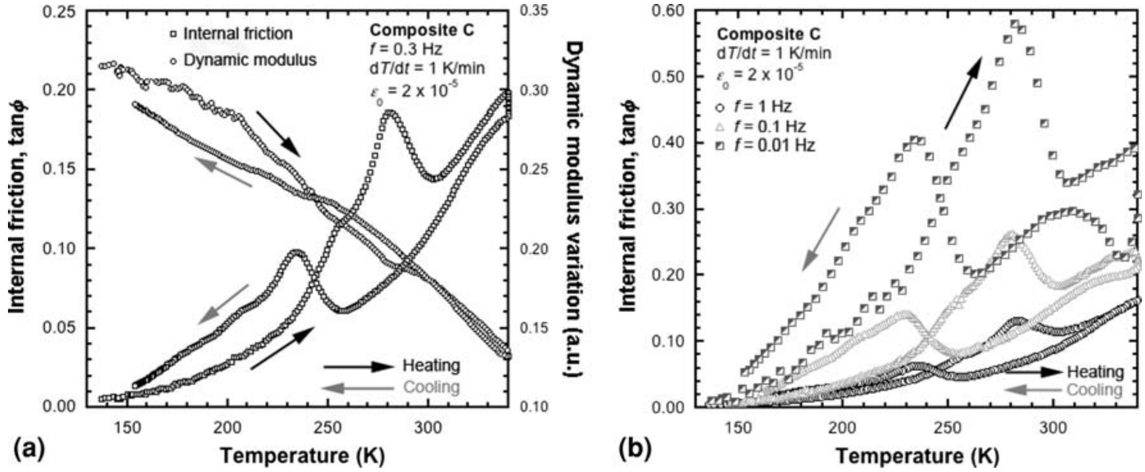

Fig. 2 (a) Internal friction spectrum and dynamic modulus variation of composite C. (b) Internal friction measurements of composite C at three different frequencies

martensite plates in the SMA particles at room temperature was not observed in composite B. This was expected because the martensitic transformation for these SMA powders takes place below room temperature (see Table 1).

Figure 1(c) shows the microstructure of composite C, where the same type of SMA particles as that used in composite B, P2, was imbedded by a eutectic In-49.1wt.%Sn matrix. The amount of SMA particles was about 60 vol.%. Even though both composites B and C were produced with the same type of SMA powders (same composition) and applying the same fabrication procedure, some differences were observed in the microstructure. As mentioned above, no particles that have undergone the martensitic transformation could be observed in composite B at room temperature; however, particles that have experienced this transformation at room temperature were clearly visible in composite C. Evidently the higher Sn content of the In-Sn eutectic matrix used in composite C has played a role in the interaction between the Cu-Al-Ni SMA particles and the molten matrix during the composite fabrication.

3.2 Damping Behavior

Figure 2(a) shows the internal friction spectra, as well as the dynamic modulus variation, of composite C for a heating/cooling cycle at a frequency of 0.3 Hz with an initial oscillation amplitude of 2×10^{-5}. It is clearly observed that there exists an internal friction peak during heating, between approximately 210 and 300 K, due to the reverse thermo-elastic martensitic transformation of the Cu-Al-Ni SMA particles of the composite, and another relatively broad peak during cooling, due to the direct transformation, between 170 and 250 K. Here, it is worth to note the very high damping observed, which reaches almost 0.2. The peaks have an associated decrease in the dynamic modulus. In addition to the peaks related to the martensitic transformation, an increasing background linked to the In-Sn eutectic matrix can also be observed when increasing temperature. The temperatures, at which the internal friction peaks are observed, are appreciably shifted toward higher temperatures with respect to the transformation temperatures of the as-atomized powders (see Table 1). This result is consistent with the microstructural observation of transformed particles at room temperature mentioned in the previous section. In addition, a broadening in the internal friction peaks was observed.

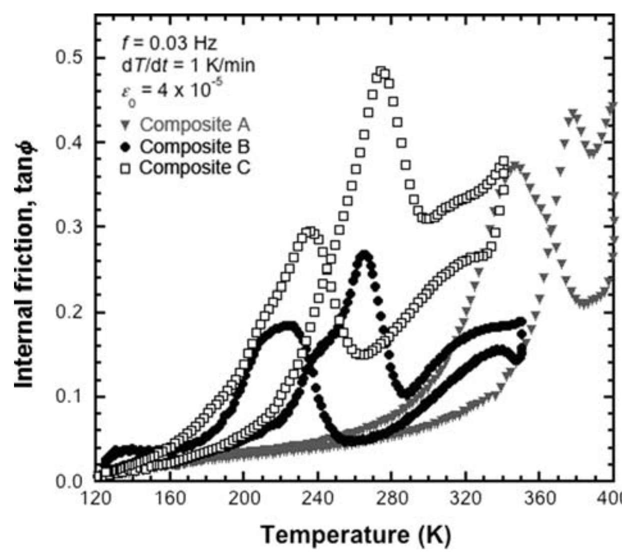

Fig. 3 Comparison of the internal friction spectra of composites A, B, and C at 0.03 Hz for an initial oscillation amplitude of 4×10^{-5}

Qualitatively, similar spectra were obtained for composites A and B (see Fig. 3 and Ref 22, 23). For composite A, the internal friction peak during heating exists between approximately 335 and 385 K, and upon cooling between 290 and 380 K. There is only a slight shift toward higher temperature, as well as a slight broadening of the transformation cycle in this composite, with respect to that of the as-atomized powders. In the case of composite B, the internal friction peaks appeared between approximately 210 and 290 K during heating, and between 170 and 250 K during cooling. These temperatures are close to the transformation temperatures of the as-atomized powders (see Table 1), although once more a certain broadening can be appreciated.

Slight temperature shifts as those observed for composites A and B are expected, because the transformation cycle in Cu-Al-Ni SMA evolves in this way upon aging (Ref 24); indeed, the powders of the composite have been aged during the elaboration process. However, even though composite C has been produced with the same powders (P2) as those used for composite B, the temperature shift in the transformation cycle was more significant. It seems that the interaction of the Cu-Al-Ni SMA particles with the higher Sn content matrix during the composite fabrication is stronger than in the other cases.

The damping of composite C for three different frequencies is shown in Fig. 2(b). As expected from the theory for the internal friction during martensitic transformations (Ref 25, 26), the lower the frequency is the higher is the internal friction exhibited by the composite. The peak position was not affected by the frequency change, which confirms that it is related to the martensitic transformation. Here, the remarkably high internal friction, reaching values higher than 0.5, is to be emphasized. Obviously, from the well-known dependence of the transitory internal friction on $\dot{T}/\omega$ (Ref 25, 26), it is expected that the damping due to the martensitic transformation of the SMA particles should be much lower in isothermal conditions, and this fact is, actually, a drawback from a practical point of view. This aspect, however, was out of the scope of the present work and only data at 1 K/min were taken.

A comparison among the internal friction spectra obtained with the three produced composites at 0.03 Hz is shown in Fig. 3. As common characteristics it can be mentioned that all composites show high damping values over a relatively broad range of temperatures. The highest damping, in a wider temperature range, has been measured for the composite produced with the In-Sn eutectic matrix, i.e. for composite C. It is evident that the higher intrinsic damping of the eutectic matrix introduces a larger background in the internal friction curves and is more attractive from a practical point of view. From this figure it can be clearly observed that a slight change in the composition leads to significant modifications in the transformation temperatures, as expected from the literature data (Ref 27). Finally, we emphasize the fact that the applied methodology enables one to match the temperature of the maximum of damping with that of work by adjusting the SMA composition. This opens new possibilities for designing high-damping materials for specific industrial or technological applications.

4. Conclusions

Composites based on Cu-Al-Ni SMAs with very high damping capacities were produced. The composite microstructures and damping properties were characterized. A good particle distribution of the Cu-Al-Ni particles in the matrices was observed. The composites exhibit very high damping capacities in relatively wide temperature ranges, values higher than 0.5 were measured. From the composites studied, the one produced with the In-Sn eutectic matrix showed the highest damping capacity in a wider temperature range. The methodology introduced provides versatility to control the temperature of maximum damping by adjusting the shape memory alloy composition.

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