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High damping capacity in porous NiTi alloy with bimodal pore architecture

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ABSTRACT

Porous NiTi exhibits pseudoelasticity, shape memory and low modulus, and thus shows application potential as high damping and bio-materials. Here $Ni_{50.5}Ti_{49.5}$ foam with bimodal pore structure was prepared by elemental powder sintering and high damping capacity demonstrated. NaCl particles of 500–600 μm and 75–90 μm diameters were adopted as space holder to create large and small pores with porosity of 40%. Incomplete densification of NiTi alloy created even smaller pores of 10–50 μm and contributed another 21.47% porosity. This NiTi foam containing bimodal pores has a compressive modulus of 1.83 GPa, smaller than that of foam with single pores or estimation by Gibson–Ashby model. This low modulus is attributed to the stress-induced martensite transformation under compressive load. The foam exhibits high damping capacities during phase transformation and at austenite region because of deformation, dislocation motion in thin nodes/walls and stress-induced martensite formation.

1. Introduction

NiTi shape memory alloys (SMAs) show a magnitude of an order of higher damping capacity than conventional metallic alloys, and attract many attentions as high damping materials in buildings, bridges [1] and high-speed trains [2]. NiTi alloy is also an appropriate material as bone tissues due to its good biocompatibility [3,4]. The high damping capacity of NiTi alloy exists during pseudoelastic hysteresis deformation [5], phase transformation and in thermoelastic martensite, the latter being related to the motion of twin boundaries, twin variants and austenite–martensite phase interfaces [1,6,7].

Damping of NiTi alloy is strain amplitude and frequency dependent [8], and is also related to prestrain and temperature [5]. Annealing of the cold rolled NiTi alloy enhances the damping capacity due to the occurrence of R-phase which softens the modulus [9]. Formation of fine Ti_3Ni_4 precipitates which have coherent interface with matrix [10] generates high stress around interface and induces stress-induced martensite, and exhibits a high transformation damping peak [10,11]. However, when the Ti_3Ni_4 precipitates grow large enough and inherent interface appears, the inner stress diminishes and damping capacity is reduced [11]. Thermal cycling produces dislocation network which hinders the austenite/martensite (A/M) transformation, making A/M damping peak diminishes in NiTi [12].

Several ways have been used to enhance the damping capacity of NiTi alloys. NiTiNb alloys with fine lamellar eutectics of NiTi (Nb) and beta-Nb dispersed in primary NiTi(Nb) phase has high damping capacity as well as yield strength [13]. Introducing graphite, which has high damping capacity, into NiTi alloy remarkably increases specific interfacial area and thus the damping property [14]. Embedding NbTi nanowires in NiTi alloy results in high energy dissipation due to the large plastic deformation in NbTi wire upon loading and interaction between NbTi and NiTi [15]. Hydrogen-doping enhances the isothermal damping capacity of NiTi alloy [16] because of the occurrence of anelastic relaxation of stress-induced motion of hydrogen-dipoles and hydrogen-twin boundary complex [17]. Solid solution elements such as Zr increase the yield strength of NiTi matrix but reduced the damping capacity [18].

It is also a feasible way to improve damping capacity of NiTi alloy by foaming strategy. The enhanced damping capacity in foams compared to dense materials was attributed to microplastic deformation, crack-initiation and motion of dislocations [19]. NiTi foams made by replication NaCl space holder show pronounced pseudoelasticity as well as shape recovery strain [20]. These porous NiTi alloys with single-modal pores also exhibit higher damping capacity than dense ones [21]. Infiltrating Mg into NiTi porous skeleton further enhances the damping capacity of NiTi foam [22]. Sun et al. [23] find that homogeneous distributed round pores resulted in higher damping capacity in NiTi foams. Other shape memory alloy foams, such as CuAlMn [24] and CuZnAl [25], also show enhanced internal friction than dense materials. So creation of porous structure as well as hydrogen-doping and incorporation of a second phase are key factors contributing to the enhancement of damping property of NiTi alloy.

Here NiTi foam with bimodal pore architecture was prepared by sintering Ni/Ti elemental powders with NaCl as space holder. NaCl space holder was chosen because of low cost, non-toxicity and low

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evaporation temperature (800 °C), which could be completely removed after sintering at 980 °C. The replacement of prealloyed NiTi [26–28] with Ni/Ti elemental powders is also cost effective. After the foam was sintered, it was solution heat treated at 1000 °C and quenched into water to reduce the content of Ni_3Ti precipitates. The pore structure and damping property of the foam were then investigated.

2. Experimental details

$Ni_{50.5}Ti_{49.5}$ foam was prepared by pressureless sintering Ni/Ti elemental powders with NaCl as space holder. Ni powders of 300 meshes, purity of 99.62% and Ti 300 meshes and purity 99.8% were used. Fine Ni and Ti powders of 5–30 μm were chosen in order to enhance the sintering. The mass ratio of Ni and Ti was 55.5:44.4 (atomic ratio of Ni/Ti was 50.5:49.5). Chemical grade NaCl particles with large (500–600 μm) and small (75–90 μm) diameters were used as space holder in order to create the foam with bimodal pores. Fig. 1 shows morphologies of the NaCl powders. The coarse powders are cubic and fine ones have irregular shape, caused by mechanical crashing.

The coarse and fine NaCl powders and Ni and Ti particles were mechanically mixed, and poured into an aluminum mold filled with acetone under stirring. The volume fraction of coarse/fine NaCl in the mixture was 40% and the ratio of coarse/fine NaCl was 73:27. For each batch of pouring, Ni/Ti/NaCl powders formed a single layer in the mold, about 0.2 mm height. This process was repeated in order that all the components homogeneously dispersed in a macro-scale in the powder mixture, similar to the layer-by-layer method used in Ref. [29]. The mixed powders were uniaxially pressed with 150 MPa and kept at this pressure for 30 min. Then the powder compact was removed out of the aluminum mold. The compact was sealed into a plastic glove and evacuated to 10^{-1} Pa, and then cold hydrostatic pressed at 200 MPa for 2 min, reaching a further volume reduction of 10%. The final compact was put into an alumina crucible and sintered in a vacuum furnace. The furnace was evacuated to 10^{-3} Pa, heated at 10 °C/min to 800 °C for 20 min, 980 °C for 5 h and 1020 °C for 3 h. The as-sintered foam was sealed into quartz tube (2×10^{-2} Pa) and heated to 1000 °C for 24 h, then quenched into room temperature water.

The porosity of the TiNi foam was determined employing Archimedes' principle. A Hitachi S-4700 SEM with EDS detector was used to investigate the pore structure and composition. A Philip's X-pert XRD was used to detect the phases with CuKα radiation at 40 kV. Compressive test was carried out in an Instron 5569 universal testing machine, with a compression rate of 0.05 mm/min. The dimension of the compressed sample was $4 \times 4 \times 8$ mm. The damping test was carried out in a Q800 dynamic mechanical analyzer (DMA) in both single cantilever (sample size $16.5 \times 5.96 \times 1.15$ mm) and tensile (sample size $10.24 \times 1.9 \times 0.5$ mm) modes. The samples were cooled from ambient temperature at 5 °C/min to −100 °C, kept at this temperature for 12 min, then heated at 5 °C/min to 200 °C. The internal friction ($\tan\delta$) and modulus (E) were recorded as a function of temperature at a constant frequency of 1 Hz and strain amplitudes of 5×10^{-4}.

3. Results and discussion

3.1. Pore architecture

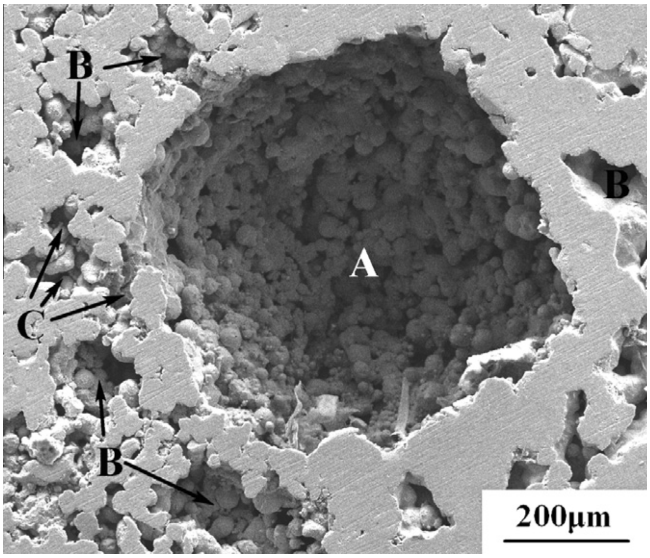

Fig. 2. SEM micrograph showing bimodal pore architecture of $Ni_{50.5}Ti_{49.5}$ foam after 1000 °C 24 h heat treatment.

Fig. 2 is the polished surface of the $Ni_{50.5}Ti_{49.5}$ foam with bimodal pore architecture. The foam contains a large pore (A) of 500–600 μm from large NaCl particle, smaller pores (B) of 75–90 μm from fine NaCl particles and irregular pores (C) from partial densification of NiTi alloy under pressureless condition. No NaCl was observed in the pores, demonstrating complete evaporation of NaCl during sintering. In fact, NaCl evaporation starts at 800 °C according to DSC results, and thus complete removal of NaCl before sintering was supposed. Accordingly, Ni/Ti compact densified without NaCl and pressure, unlike the liquid NaCl assist sintering by hot hydrostatic sintering [30]. The foam shows good bonding state after sintering at 980 °C 5 h following 1020 °C 3 h. This good sintering state is achieved by cold hydrostatic pressing at 200 MPa resulting in well contact of Ni/Ti powders before sintering, and by transient liquid sintering at 1020 °C (much higher than liquid appearance temperature of 942 °C).

The first stage sintering temperature of 980 °C is higher than commonly accepted sintering temperature (942 °C) of NiTi. Cluff et al. [31] found that Ti_2Ni, NiTi and Ni_3Ti formed when sintered at 942 °C; above 942 °C, liquid phase appeared and might induce combustion reaction, which resulted in the reduction of density. In order to increase the density, two stages sintering profile [32–35], that is sintering at low (stage I: ~942 °C) and high (stage II: 1000–1100 °C) temperatures, were carried out. Here, a two-stage profile was employed to improve densification of Ni/Ti powders, but at a higher stage I temperature of 980 °C in order to accelerate the diffusion and reaction process.

Fig. 3 shows 3-D morphologies of the $Ni_{50.5}Ti_{49.5}$ foam with 61.47% porosity. From Fig. 3(a), the foam structure may be described as semi-close pore structure with big nodes connecting walls. Fig. 3(a) shows big pores (A) distributed homogeneously, with loose walls surrounding them. The size of these big pores is 500–600 μm, replicating the shape and size of large NaCl powders. The walls contain many small pores and a few big "windows"

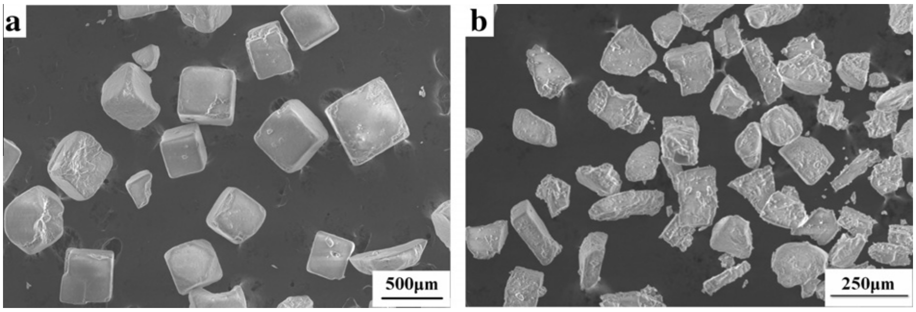

Fig. 1. SEM micrographs showing morphologies of coarse (a) and fine (b) NaCl space holders. The coarse NaCl powders with diameter of 500–600 μm are cubic shape, and fine NaCl powders with diameter of 75–90 μm are irregular shape.