AI-NiTi METAL MATRIX COMPOSITES FOR ZERO CTE MATERIALS: FABRICATION, DESIGN, AND MODELING

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Abstract

Al-NiTi composites fabricated via ultrasonic additive manufacturing (UAM) provide a light-weight solution for low thermal expansion applications. It is shown that the thermal expansion of Al 6061 can be reduced by over 50% by incorporating a 13% volume fraction of NiTi fibers. This reduction in thermal expansion occurs from the contraction of the NiTi fiber during heating, thereby offsetting the thermal expansion of the Al matrix. Al-NiTi composites are made possible by low temperature UAM process. Successful implementation of these composites requires a careful design approach that includes the processing characteristics as well as the thermo-mechanical response of the shape memory fibers and matrix. This is achieved using a NiTi microstructure based FEA model implemented that captures the underlying thermomechanical response of the NiTi fibers and calculates the complex stress state within the composite.
Introduction

Ultrasonic additive manufacturing (UAM) is a solid-state consolidation process where thin foils of similar or dissimilar metals are ultrasonically joined in a layer-by-layer process to form gapless, 3D metal parts [1]. Periodic machining is utilized during the UAM process to achieve complex designs, features, and to embed various objects into the structure, such as reinforcing fibers. A schematic of the UAM process is shown in Figure 1. Due to the physics of ultrasonic welding, metallic bonding takes place at temperatures far below metallic melting temperatures [2]. Thus, temperature-sensitive metals or components such as nickel titanium (NiTi) shape memory alloys can be integrated into metallic structures. The weld power of current UAM systems has increased nearly an order of magnitude over the past five years. The increase in weld power and available down force remedies poor interfacial properties observed in early UAM builds [3].

Figure 1: UAM process for developing low temperature metal-matrix composites.

The low formation temperature and additive nature of UAM enables the manufacture of NiTi-Al composites. NiTi fibers can be integrated into the Al matrix to counteract thermal expansion upon heating. Specifically, when the composite is heated, the strain recovery of the NiTi fibers counters the expansion of the aluminum matrix. This combination creates a light, stiff, and thermally-stable material for engineering applications. Previous efforts have shown a 60% reduction in the average coefficient of thermal expansion for Al 3003 up to 100°C [4]. Yet, metallic bonding between the aluminum matrix and NiTi fibers was not always observed. Instead, the interface was believed to be predominately supported by mechanical coupling in the form of a friction fit since the preexisting oxide layer on the fiber was still present after welding [5]. The bonding between the Al matrix and NiTi fibers was improved by removing the preexisting oxide layer; the results were evaluated tested in a recent interface study using Al 6061 [6]. Fiber pullout testing and Energy Dispersion X-ray Spectroscopy (EDS) were used to test the Al-NiTi interface strength and bond character, respectively. Fiber pullout testing showed that the matrix failed for all surface finishes under investigation, including those with the preexisting oxide. On the other hand, EDS supported the possibility of metallic bonding between the Al matrix and NiTi fibers when the preexisting oxide was removed prior to welding since little to no oxide concentration was observed at the NiTi-Al interface after welding. Lastly, this study showed that circular fiber geometries exhibited strong con-
solidation character because the matrix was able to flow in and around the fibers during fabrication. The resulting interface exhibited strong resistance to ply delamination.

The focus of this paper is to test these recent NiTi fiber consolidation and bonding observations and to develop an accurate model that describes composite behavior. Multi-layer NiTi-Al composites were fabricated and tested with Al 6061 and mechanically polished NiTi fibers. The corresponding simulations utilized a microstructural constitutive model of NiTi implemented in ABAQUS. This microstructural model describes underlying NiTi physics, i.e., texture effects, martensite variant interaction, and crystal plasticity. Overall, it can simulate a variety of engineering materials and predict the stress redistribution and local thermo-elastic response within the composite. Previous models of NiTi-Al composites utilized a phenomenological approach implemented in MATLAB [7]. Although they require little computational power, these models captured only the axial stress state of the NiTi fiber and did not predict the interface or shear stress. The simulations reported in this work are calibrated to single fiber mechanical tests using a crystallographic description of the transformation process [8].

Methods

Material Stabilization (Training) and Characterization

Mechanically polished 0.381 mm (0.015") diameter NiTi fibers were obtained from Nitinol Devices & Components, Inc. for this study. The fibers had an austenite start temperature above room temperature, were trained to be straight. A stabilization process was applied to the fibers prior to composite construction [9] in order to minimize changes in flow stress and inelastic strain when the NiTi fibers were incorporated into the Al matrix. The stabilization process involved cycling the material 10 times at 75°C to ensure that the material was in the super-elastic state, i.e., above the Austenite finish temperature. These cycles were performed using a TestResources 131R1000-6 tensile frame with MTS Screw Action Grips with serrated faces and a thermal chamber.

The stabilized NiTi fibers were then characterized using isothermal tensile testing and differential scanning calorimetry (DSC). Isothermal cyclic tensile tests were performed on a 158.8 mm (6.25 in) long piece of NiTi fiber at 65°C, 75°C, and 85°C to quantify the stress as a function of temperature. The nominal strain rate was 0.8%/min to minimize internal heating during the test in accordance with ASTM F2516, Standard Test Method for Tension Testing of Nickel-Titanium Super-Elastic Materials. The nominal strain was determined by dividing measured displacement by the fiber’s length.

The austenitic transformation temperatures were measured using a TA Instruments 2920 Differential Scanning Calorimeter. The sample was first cooled to near −10°C to ensure that the NiTi fiber was in the martensitic state or below the martensite finish (\(M_f\)) temperature. The sample was then heated to 100°C at 10°C/min while simultaneously measuring the reference heat flow. The transformation temperatures were determined by fitting lines to both sides of the transformation peak and finding the intersections between the plateaus as specified in ASTM F2004, Standard Test Method for Transformation Temperature of Nickel-Titanium Alloys by Thermal Analysis. This DSC system was not used to measure the martensitic transformation temperatures because reliable results for cooling could not always be achieved.
Composite Design, Fabrication, and Testing

A 9 kW UAM system was utilized (Figure 2). The metal matrix used in this study was Al 6061-H18. Foils were 0.152 mm (0.006 in) in thickness and 25.4 mm (1 in) in width. Aluminum 6061 was chosen for its high strength to weight ratio and close compatibility with UAM.

Figure 2: State-of-the-art UAM system fitted with a 9 kW weld head (inset), a 3-axis CNC mill with 25 HP, 8000 RPM spindle, and integrated 40 W laser.

The composite design for this study placed NiTi fibers away from shared interface boundaries to minimize delamination risk. The upper and lower portion of a fiber did not share the same common boundary with the upper and lower portions of a neighboring fiber, see Figure 3(a). A 0.397 mm (0.016 in) diameter ball-nose end mill was used to cut a 0.356 mm (0.014 in) deep channel in the Al matrix to assist in the encapsulation and placement of the NiTi fibers during consolidation. The channel dimension was 0.025 mm (0.001 in) less than the NiTi fiber diameter to enhance scrubbing and plastic flow of the Al around the fiber and into the pocket, Figure 3(b). Welding was conducted on 101.6 mm × 76.2 mm × 4.76 mm (4.0 in × 3.0 in × 0.188 in) Al 6061-T6 build plates that were constrained in place using a custom base plate, fabrication fixture, and vacuum chuck. Table 1 shows the parameters that were used to weld the aluminum tapes. These parameters were obtained from a recent design of experiments for optimal welding parameters of Al 6061-H18 [10] and they have been shown to be sufficient in pilot NiTi-Al welds.

1Al 6061-H18 was fabricated by work hardening annealed 6061 material
Figure 3: Composite design: (a) fibers were strategically placed to eliminate shared interface boundaries and minimize possible delamination; (b) pocket design for fiber placement.

Table 1: Ultrasonic welding parameters used in study.

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Value</th>
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</thead>
<tbody>
<tr>
<td>Temperature</td>
<td>22°C (70°F)</td>
</tr>
<tr>
<td>Force</td>
<td>6000 N</td>
</tr>
<tr>
<td>Amplitude</td>
<td>34.6 μm (74%)</td>
</tr>
<tr>
<td>Speed</td>
<td>84.6 mm/sec (200 in/min)</td>
</tr>
</tbody>
</table>

Removal of the completed composite from the surface of the build plate was done in multiple steps: (i) CNC milling to reduce the composite geometry to the desired dimensions, (ii) manual milling to remove the build plate from under the composite, and (iii) a slow speed cutting with a high volume coolant flow to remove the end of the composite attached to the remaining build plate without significant heat generation. Two composites were fabricated. Each measured approximately 5.49 mm × 74.00 mm × 1.27 mm (0.22” × 2.91” × 0.05”). The NiTi volume fractions for the two composites were 13% and 13.8% respectively. Slight manufacturing variations in the removal process caused the volume fraction to be slightly different.
Thermal testing was performed on the composites using a custom 15.24 cm × 15.24 cm (6.0” × 6.0”) thermal chamber with foil-faced fiberglass insulation, Figure 4. The top of the chamber was equipped with a circular viewing window. A Milwaukee Variable Temperature Heat Gun supplied heated air to the chamber, and this air was deflected upon entering the chamber with an aluminum baffle to produce even heating of the samples. Composite CTE measurements were made by mounting strain gauges on a reference piece of UAM Al 6061 and composite so that the relative strain difference between the two materials could be determined [11]. The composite and the reference sample were mounted within the chamber in a free-stress, cantilevered condition. Micro Measurement WK 13031CF350 strain gages were mounted to the top and bottom sides of both samples, approximately 3.8 cm (1.5”) from the fixed end. The temperatures of each sample were measured using type-J thermocouples with Omegatherm 201 thermal conductivity paste at the composite thermocouple interface to ensure accurate measurements. The outputs of the strain gauges and thermocouples were connected to a National Instruments data acquisition system, which interfaced with a LabVIEW VI for recording data. Before each sample was tested, the strains were zeroed using the calibration features within LabVIEW. Each test was performed by heating the sample from room temperature to approximately 100°C using the slowest speed setting on the heat gun (near 45 min. to reach 100°C). Samples were then cooled back to room temperature by setting the heat gun to 30°C and allowing the chamber to equilibrate with the room once the heat gun could no longer provide cooling (near 1 hr).

Composite Modeling

Numerical simulations of the composite were conducted to better understand the behavior of the fabricated Al-NiTi composite, especially the matrix-fiber interaction that determines its performance. The primary challenges in modeling are to accurately capture the response of isolated NiTi-fibers and to capture the physics of stress redistribution within the fiber-matrix composite. A microstructural finite element constitutive model [8] was utilized to meet both challenges. This model is implemented as a User-defined MA-
The NiTi fiber response was calibrated to the stress-strain response at three different temperatures and the critical transformation temperatures determined from Differential Scanning Calorimetry (DSC). The full list of calibrated parameters of the NiTi model is summarized in Table 2. The phase transformation between austenite and 24 correspondence variant pairs (CVPs, a.k.a., martensite plates) is controlled by the driving force per unit volume described by

\[ f_t = b_t \cdot \left( (F^e)^T F^e T^* \right) m_t - \frac{\lambda_T}{\Theta_T - \Theta_T} - \sum_{u=1}^{N_T} h_{tu} v_u \]

where \( t = 1 \) to 24 denotes the martensite plate type; \( b_t \) and \( m_t \) are the shear magnitude and shear plane normal vectors associated with the transformation from austenite to martensite plate type \( t \), \( T^* \) is the 2nd Piola-Kirchhoff stress, and \( F^e \) is elastic part of the deformation gradient; \( \lambda_T \) is the transformation latent heat, \( \Theta_T \) is the thermodynamic equilibrium temperature, \( \Theta \) is the current temperature; \( v_u \) is the volume fraction of plate type \( u \), and the \( h_{tu} \) matrix specifies the interaction energy among the plates. In terms of physical significance, the first term on the right-hand side expresses the mechanical contribution to the driving force of transformation from austenite to CVP-type \( t \); the second term expresses the thermal contribution; and the third term expresses the inter-CVP interaction. Evolution of volume fraction of each martensite CVP \( (v_t) \) is controlled by the following consistency condition

\[-f_c \leq f_t \leq f_c\]

where \( f_c \) is the critical force/unit volume for transformation. This will be determined through calibration. For further details regarding the development and implementation of the SMA model, the readers are referred to Ref. [13] and [8].

Table 2: Material parameters of the NiTi fiber (calibrated) and Al matrix (from [4]).

<table>
<thead>
<tr>
<th>Category</th>
<th>Parameters and their calibrated/adopted values</th>
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<tbody>
<tr>
<td>Elastic and thermal properties of the fiber</td>
<td>Elastic constants of austenite (GPa): ( C_{11}(A) = 130; \ C_{12}(A) = 98; \ C_{44}(A) = 34 ); Elastic constants of martensite (GPa): Same as austenite; Thermal expansion coefficients ( (10^{-6}/K) ): ( \alpha_A = 11; \ \alpha_M = 6.6 )</td>
</tr>
<tr>
<td>Phase transformation properties of the fiber</td>
<td>Equilibrium transformation temperature (K): ( \Theta_T = 303 ); Latent heat of transf./unit vol. (J/cm³): ( \lambda_T = 151 ); Hardening matrix ( h: h_{tu} = \begin{cases} C_{44(A)}/1500 \ (t = u) \ C_{44(A)}/3000 \ (t \neq u) \end{cases} ); Critical driving force force (J/cm³): ( f_c = 6.0 ); Number of martensite CVPs: ( N_T = 24 ); Max. principal shear strain by one CVP: (</td>
</tr>
<tr>
<td>Elastic and thermal properties of Al</td>
<td>Young’s modulus (GPa): ( E_{Al} = 68.0 ); Poisson’s ratio: ( \nu_{Al} = 0.33 ); Thermal expansion coefficients ( (10^{-6}/K) ): ( \alpha_{Al} = 23.6 )</td>
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</table>
Since the matrix-fiber interface is expected to yield or fracture much earlier than either of the pure phases, plastic deformation of the fiber is not incorporated into the modeling in this work. This is accomplished by setting the reference plastic slip rate to zero in the NiTi model. The fiber is meshed with eight-node, coupled-temperature-displacement, reduced-integration elements (C3D8RT) in the ABAQUS element library. Each element is treated as a separate austenite grain, and assigned a crystallographic orientation within a random variation of magnitude of ±5° from [110]B2, approximating the RD texture reported for the same Nitinol product in Ref. [14], as suggested by the manufacturer. A total of 800 C3D8RT elements (and therefore, grains) are used for the fiber. The texture data used in the simulation has been found to significantly influence the predicted composite performance, a result in agreement with previous works [15].

In the composite simulation, only one-eighth of a fiber-containing Representative Volume Element (RVE) is modeled by applying a mirror-plane condition to the $x = 0$, $y = 0$ and the $z = 0$ planes. Comparison with the full model showed that no significant difference results from such a simplification. Figure 5 shows the RVE, which contains 13 vol.% of fiber (shaded part). Considering the mirror plane perpendicular to the fiber axis, this RVE corresponds to a length-to-diameter ratio of 20:1, which is in agreement with the experimental geometry. The aluminum matrix is modeled as purely elastic, with properties given in Table 2. To further simplify the computation, a uniform temperature over the whole model is assumed at any given instant so that heat conduction within the sample is not explicitly modeled. Also, the matrix-fiber bonding is assumed to be perfect, i.e., no-slip.

Figure 5: The finite element model of the composite, using one-eighth of a fiber-containing RVE. Three mirror-plane conditions are applied – two of them through the center axis of the fiber (shown in blue), while the remainder is perpendicular to the fiber.
Figure 6: Simulated mechanical and thermal loadings: the NiTi fiber is first cooled under a tensile stress of $\sigma_{\text{bias}} = 400$ MPa in the absence of Al matrix to induce oriented martensite, and then unloaded at 22°C (blue regime). This is followed by embedding the fiber in matrix at $t = 500s$. The heating-cooling cycle of the composite from 22°C to 100°C is repeated 4 times (orange regime) to study the evolution of performance.

Two simulation efforts are documented here. The first one is the calibration of NiTi fiber parameters against isothermal tensile loading-unloading curves at three different temperatures: 65°C, 75°C and 85°C, as well as the DSC curve measured during heating of the NiTi fiber at its pre-deployment state. This is followed by simulations of the composite at working condition. The double Y-axis plot of Fig. 6 shows the loading history of the deployment (blue regime) and that of the subsequent thermal cycles (orange regime). A total of four thermal cycles between 22°C and 100°C are simulated.

Results

NiTi Fiber Performance

Characterization results from isothermal tensile testing and DSC are shown in Figure 7. It should be noted that, as described in the Methods section, the behavior shown in Figure 7(a) is achieved after mechanically cycling the NiTi fiber multiple times in a super-elastic state, thereby stabilizing the response of the as-received material. This is due to the crystallographic defects induced during the stabilization process, such as transformation-induced dislocations [16]. Since the focus of this work is the composite behavior, the NiTi model is calibrated to the stabilized behavior. The resulting parameters are listed in Table 2. As shown in Figure 7, the calibrated NiTi fiber model captures both the uniaxial tension and DSC behavior reasonably well.
Figure 7: Model calibration: (a) against isothermal stress-strain curves; (b) against DSC

Experimental Composite Performance

Figure 8 shows that empirical strain-temperature results for both NiTi-Al composites and pure UAM Al-6061. The composites begin transforming near 40°C and exhibit compression from the NiTi. The first cycle demonstrates enhanced compression and curvature during heating, which may be linked to the collapse of voids within the UAM build. Lastly, the NiTi-Al composites exhibit CTEs less than 50% of Al-6061.

Figure 9 shows the optical images of the end of the composite before and after heating integrity of the fiber-matrix interface. The composite end was imaged to assess the shear transfer taking place between the NiTi fiber and Al matrix. Damage in the composite due to testing is likely to initiate at the composite end. The image is inverted since the image is not of the same sample but is of the same cross section, i.e., some material was removed between samples with a saw. As a result, key fibers have been boxed. Some voids are present in the composite prior to testing, yet no obvious cracking can be seen. After testing, the composite remains reasonably crack free.
Figure 8: Empirical strain-temperature results: (a) 13% NiTi volume fraction composite for 3 cycles; (b) comparison of Al (no NiTi), composite with 13% volume fraction and composite with 13.8% volume fraction. Fit slopes from 40 to 100°C are given.

Figure 9: End cross section of 13.8% NiTi composite: (a) the composite before heating, some void presence can be seen with no obvious cracking; (b) the composite after 3 heating cycles, little to no cracking formed during thermal cycling.

Simulated Composite Performance

The composite strain-temperature response from the FE simulations is shown in Figure 10(a). The composite strain is measured and averaged over the middle cross-section plane of the matrix, thus avoiding end effects (or St. Venant effects, Figure 11).
Figure 10: FE simulation results: (a) Strain vs. temperature curve for the first 4 cycles, of which the latter three effectively overlap; (b) Evolution of martensite volume fraction in the first 4 cycles.

Figure 11: Axial normal stress ($\sigma_{zz}$) distribution at $T = 100^\circ$C during the 1st cycle. The highlighted nodes show the plane on which the composite strain is measured. It can be seen that it is far from the St. Venant region near the end of fiber-matrix interface.

Several observations can be made regarding the comparison between the simulation result in Figure 10(a) and experimental result in Figure 8(a). First, the pronounced difference between the 1st-cycle and subsequent-cycle behaviors is partially captured in the FE simulation, the reason of which will be discuss later. Second, the transition in apparent CTE of the composite in the temperature regime of 40 - 60°C is also captured for subsequent cycles. This corresponds to the onset of the martensite-to-austenite transformation, as shown in Figure 10(b). Third, the contraction above 70°C in Figure 8(a) is absent in Figure 10(a), which implies that the FE model does not capture these physics. Fourth, the simulation slightly under-predicts the maximum strain at 100°C, under-predicts the strain hysteresis for the 1st cycle, and over-predict the strain hysteresis for “subsequent” cycles. These discrepancies are likely interconnected, as will be discussed in the next section.
Discussion

Evolution of Fiber State

The thermo-mechanical mechanisms underlying the composite behavior are discussed first. The driving force for the change of CTE of the composite arises from axial contraction of the NiTi fiber. This contraction arises from the phase transformation from an oriented martensite (B19') phase to a high temperature austenite (B2) phase. Constrained by the Al matrix, this phase transformation competes with the conventional thermal expansion of the Al matrix as the composite is heated. This generates an internal stress field that reduces the amount of expansion in the Al matrix and, at the same time, limits the shape recovery of the NiTi fiber.

The evolution of the NiTi fiber state is schematically shown in Figure 12. The fiber starts at the fully reoriented martensite state (Point A) at the beginning of the 1st-cycle. During heating, it enters in the martensite-to-austenite (M-A) transformation region, but the transformation is only partial (to Point B), as revealed by the martensite volume fraction of 82% at $T = 100^\circ$C in Figure 10(b). As the sample is cooled from 100$^\circ$C, the martensitic (A-M) transformation occurs. However, it is again incomplete by the end of the cooling, achieving only 95% martensite. Therefore, the fiber still lies in the A-M (blue) region at the end of the 1st cycle, as represented by Point C in Figure 12.

Figure 12: Stress-Temperature phase diagram of the NiTi fiber. The upper (blue) region corresponds to the austenite-to-martensite (A-M) transformation region, and the lower (red) region corresponds to the martensite-to-austenite (M-A) transformation region. No transformation happens in the region between these, and the volume fraction of phases here totally depends on previous path. The 1st-cycle trajectory shown here is based on the present finite element simulations.

The partial A-M transformation is found to be the mechanism behind the compressive strain at the end of the 1st cycle shown in Figure 10(a). It arises from the difference in
A-M mechanical driving forces during the reorientation stage (blue regime in Figure 6) and that during in situ cooling in the composite. Since this difference is eliminated with the completion of the 1st cycle, subsequent cycles shift the fiber state between Point C and Point B in Figure 12, therefore they overlap in the strain-temperature curve in Figure 10(a) and nearly overlap in Figure 8(a).

**Not-captured Experimental Features**

The largest difference between experiment (Figure 8(a)) and simulation (Figure 10(a)) is the contraction at the high temperature end of the 1st cycle. The mechanisms considered in the FE model does not capture this feature. Since there is experimental evidence that the voids in matrix are generated in the UAM process, it is possible that collapse of these voids may be responsible for this behavior. This hypothesis agrees with three known facts: i) The level of internal stress in the composite increases sharply with temperature at $T > 50^\circ$C. Consequently, the contraction is observed at the most stressed state and agrees with void collapse; ii) The contraction does not occur in subsequent cycles. iii) The compressive strain at the end of the 1st cycle is significantly larger in the experiments (Figure 8a) compared simulations (Figure 10a). All of these support the hypothesis that the aluminum matrix plastically deforms through the collapse of voids.

To further evaluate the void collapse hypothesis, the 13.8% NiTi composite was cross sectioned and polished at the strain gauge location to examine void presence. Cross sections before and after testing is shown in Figure 13. The significant reduction in void volume fraction from thermal cycling supports the collapse-of-void hypothesis.

![Figure 13: Comparison of end and strain gauge cross sections of the 13.8% NiTi composite: (a) the composite before heating, with significant voids; (b) the composite after 3 heating cycles, showing reduced void volume fraction.](image)

The collapse of voids and axial plastic contraction of the matrix during the first cycle may explain the under-estimation of composite strain and over-estimate of strain hysteresis in subsequent cycles. The decrease in the natural length of the matrix during the first cycle will make the axial stress in fiber lower than in the FE simulations (~110MPa). In Figure 12, a lower stress than that at Point C will reduce the volume fraction of stress-induced martensite volume fraction at 22$^\circ$C to less than the 95% predicted value shown in Figure 10b). The reduced martensite volume fraction will decrease the available martensite for transformation. Since the ability of contraction in fiber is determined
by the amount of transformation that occurs during a thermal cycle, a lower martensite fraction at 22°C means lower constraint on the thermal expansion of the Al matrix. This can explain why the observed strain in subsequent cycles is higher in the experiments compared to simulations. Also, a lower martensite fraction at room temperature implies that transformation during subsequent thermal cycles is also lower, which agrees with a smaller strain hysteresis.

**Conclusions**

NiTi-Al composites provide a light weight alternative compared to iron-based thermally invariant materials. This was demonstrated by fabricating Al 6061 matrix composites with approximately 14 vol% aligned commercially-available NiTi wires. The composite exhibited a reduced coefficient of thermal expansion of 11.1 $\mu \varepsilon/^\circ C$ over the range 45-100 C, compared to 23.6 $\mu \varepsilon/^\circ C$ for the matrix. A state-of-the-art microstructural NiTi model implemented in ABAQUS was developed to capture key composite and NiTi physics. The simulations were calibrated to single fiber stress-strain and differential calorimetry experimental data. The predicted composite response qualitatively captured the experimental results, including an initial axial contraction of the composite $10^{-4}$ during the first thermal cycle and a reduced coefficient of thermal expansion during subsequent cycles. Differences between the experiments and simulations arise from the model assumption that the matrix does not plastically deform. However, experimental observations show the collapse of voids around fibers during thermal cycling. The results suggest that the Al matrix plastically deforms during initial thermal cycling and then stabilizes. In principle, the simulations can incorporate plastic deformation and work hardening of the matrix. Overall, this combined experiment-simulation approach can be used to accelerate the development of thermally invariant, shape memory based composites.

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