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Single phase boundary actuation of a ferromagnetic shape memory foil

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Abstract

The novel mechanism of temperature-gradient-induced single phase boundary actuation is presented for a single crystalline ferromagnetic shape memory alloy (FSMA) foil. It is shown that applying a temperature gradient along the FSMA foil specimen results in the formation and propagation of a martensite–austenite phase boundary from the hot to the cold side, allowing for reproducible strain–temperature characteristics. The selection of martensite variants upon phase transformation is controlled by simultaneously applying a bias magnetic field, which determines the maximum strain response. Single phase boundary actuation is demonstrated for a Ni–Mn–Ga foil of 100 μm thickness with 10 M martensite structure at room temperature. A small temperature gradient of 5 K mm^{-1} and a bias field along the temperature gradient of 120 mT are sufficient to achieve the maximum possible strain of 4.1%, corresponding to the length difference of the short *c*-axis of tetragonal martensite and the axis of cubic austenite. For a bias magnetic field in the perpendicular direction, the maximum strain change is −1.9%.

Keywords: Martensitic phase transformation; Phase boundary; Magnetic shape memory alloys; Ni_2MnGa; Digital image correlation

1. Introduction

In recent years, the effect of magnetic-field-induced reorientation (MIR) in ferromagnetic shape memory alloys (FSMAs) has been studied extensively [1–4]. In particular, near-stoichiometric Ni–Mn–Ga single crystals have been the focus of both fundamental and applied research due to the particularly large MIR effect, of the order of 10%, which is of great interest for actuation and sensing applications [2,3]. The large strain is ascribed to the reorientation of martensitic variants that are separated by mobile twin boundaries (TBs) [2]. Exploitation of this effect, however, encounters a number of drawbacks, including (i) the existence of a critical twinning stress and corresponding critical magnetic field that needs to be overcome to induce MIR, (ii) the low reproducibility of strain response during partial MIR and (iii) the relatively low blocking stress that limits the available work output [4].

The first two drawbacks are related to the types of coexisting twins that may counteract each other, as well as to surface defects and internal pinning sites causing non-monotonic TB motion [5,6]. The non-monotonic TB motion can also be explained by the periodic energy barriers that dominate the quasi-static motion of the TB. In particular, it was found that the typical distance between adjacent barriers is ∼19 μm, while their effective amplitude is ∼0.16 J m^{-2} [7].

In small-scale applications, the influence of surface defects becomes increasingly severe due to technological constraints in fabrication of FSMA foils and films. In order to reduce the twinning stress and improve the TB mobility, a number of training methods have been developed by combining thermal, mechanical and magnetic loading [8]. Compared to bulk, FSMA films and foils encounter the special constraint that mechanical training by compression in the actuation (length) direction is hardly possible as the

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large length-to-thickness ratio results in buckling. For an FSMA foil of 200 μm thickness, it has been shown that thermomechanical training by multiple heating and cooling the foil through the transformation temperature regime under tensile load allows us to preferentially align the long *a*-axis along the tensile direction and to reduce the twinning stress from 5 down to 0.3 MPa [9].

The drawback of low blocking stress is related to the magnetocrystalline anisotropy energy of the material. A common approach to optimizing the magnetocrystalline anisotropy energy is to adjust the chemical composition such that the difference between martensitic transformation temperature M_s and Curie temperature T_c is large. Selecting an operating temperature T_o near M_s, allows for low twinning stress at the same time [10]. Recent work showed some evidence that size reduction may lead to an increase of blocking stress due to energetical preference of selected twin variants, which enhances the energy barrier of magnetization rotation [11].

Here, we propose and explore actuation of FSMA foil specimens subjected to an adjustable thermal field gradient and a bias magnetic field as a means to control the formation and propagation of a martensite–austenite phase boundary. The main motivation is to overcome the drawbacks of limited blocking stress and low TB mobility in state-of-the-art FSMA film and foil specimens and, thus, to open up a route to realize FSMA foil/film actuation with large work output at low magnetic field. Despite the use of thermal actuation, response times are expected to be still attractive for applications due to the large surface-to-volume ratio.

2. Materials properties

An ingot of single crystalline bulk Ni–Mn–Ga has been grown by AdaptaMat Ltd. The chemical composition is determined by energy-dispersive X-ray (EDX) analysis to be $Ni_{50.2}Mn_{28.4}Ga_{21.4}$ (at.%). Details on the preparation of the bulk samples can be found in Ref. [12]. The Ni–Mn–Ga foils are cut from bulk single crystals along the [1 0 0] direction of the Ni–Mn–Ga unit cell by wire sawing. The deformed surface layer is removed by electropolishing. The final foil thickness is adjusted to be 100 and 200 μm. Stripe specimens are cut to obtain lateral dimensions of 12 × 3 mm^2 with orientation of the long edge in the [1 0 0] direction. X-ray diffraction measurements show that as-prepared Ni–Mn–Ga stripes have a 10 M martensite structure at room temperature [13]. The start and finish temperatures of the martensitic transformation, M_s and M_f, and of the corresponding reverse transformation, A_s and A_f, are investigated by temperature-dependent electrical resistance measurements. Upon heating and cooling, abrupt changes of electrical resistance are observed in the phase transformation regime, as shown in Fig. 1. The transformation temperatures M_s and M_f are determined to be 52 and 50 °C, respectively, while A_s and A_f are 57 and 58 °C, respectively. The Curie temperature is

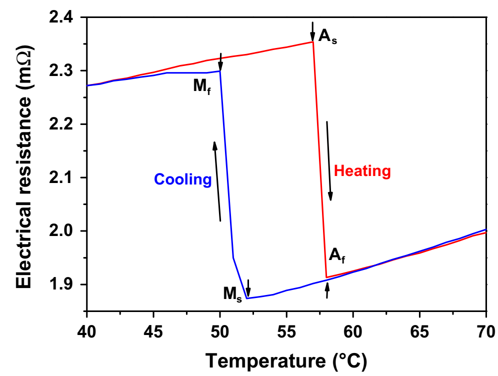

Fig. 1. Electrical resistance characteristics upon heating and cooling for a Ni–Mn–Ga stripe specimen of 100 μm thickness. The temperatures of martensitic and reverse transformation are indicated.

determined by differential scanning calorimetry to be 99 °C. The corresponding measurements can be found in the Supplemental Information.

Fig. 2 shows typical magnetostrain characteristics of a Ni–Mn–Ga stripe specimen of 200 μm thickness at room temperature (20 °C) and 50 °C. One end of the specimen is fixed to a ceramic substrate, while the other end is kept free-standing. The magnetic field is ramped at a rate of 2 mT s^{-1}. At room temperature, the critical magnetic field for MIR is ∼160 mT. Upon further increase of the magnetic field, a typical stepwise increase of magnetostrain is observed, reflecting the jump-like propagation of twin boundaries caused by intermittent pinning at internal obstacles and surface defects [6]. The reorientation process completes at ∼0.24 T. The maximum magnetostrain is ∼5.3%, which is below the limit of ∼6% given by the tetragonal distortion in 10 M martensite. The reduced strain is due to inactive regions near the fixation of the specimen, which do not contribute to reorientation [14]. When decreasing the magnetic field, the magnetostrain remains unchanged, reflecting the metastable nature of martensite. Resetting of the magnetostrain is achieved by applying

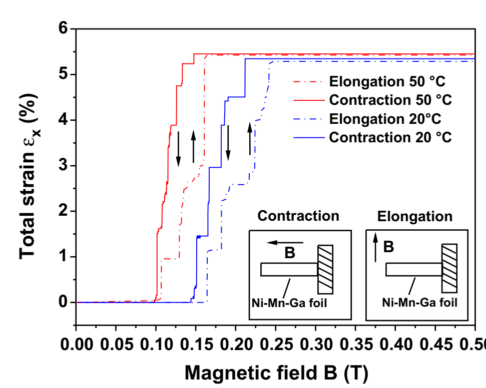

Fig. 2. Magnetostrain characteristics in the *x*-direction of a Ni–Mn–Ga stripe specimen of 100 μm thickness at room temperature (20 °C) and 50 °C for magnetic fields applied in longitudinal and transverse directions, as sketched in the inset.