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# Twinning in Ni–Fe–Ga–Co shape memory alloy: Temperature scaling beyond the Seeger model

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ABSTRACT

The temperature dependence of the macroscopic de-twining stress in Ni–Fe–Ga–Co shape memory alloy is proportional to the temperature dependence of the micro-yield stress. This proportionality holds for all values of the reversible anelastic strain derived from non-linear internal friction experiments and contradicts predictions of the Seeger model for the temperature dependence of the yield stress. The Seeger model requires two length scales which leads to the independence of two yield stress components. Instead, we find the self-affine behaviour of anelastic strain, which excludes all separations of lengthscales.


The origin of the temperature dependence of de-twinning stress in magnetic martensites is not understood [1–9]. It seems agreed, however, that the de-twinning stress and its temperature dependence in ferromagnetic and metamagnetic shape memory alloys determine the actuation by a magnetic field [1,2,3,6,10]. The temperature dependence of the de-twinning stress seems to reflect thermal activation, while we have no clear atomistic model that explains such activation process. Heczko and Straka observed a strong exponential-like increase of the twinning stress between 310 and 120 K in five-layered Ni–Mn–Ga martensite [1]. Straka et al. [2] attributed the effect of temperature on field-induced motion of a single twin boundary in a Ni–Mn–Ga 10 M martensite to thermally activated motion of twinning dislocations. For a similar 10 M Ni–Mn–Ga alloy with polyvariant structure, Peng et al., using internal friction data, found the activation energy of twin boundary motion of 0.6 eV [4]. Okamoto et al. [3] reported a nearly linear temperature dependence of stress required for twinning plane movement in Ni–Mn–Ga. Sozinov et al. observed an intermediate as compared to type I and type II twins effect of temperature on de-twinning stress for compound twins in non-modulated Ni–Mn–Ga–Co–Cu alloy [10]. Panchenko et al. mentioned a weak, likely, linear temperature dependence of the de-twinning stress in Ni–Fe–Ga–Co down to 77 K [11].

Estimates of activation energies related to the twin boundary motion from Arrhenius-type relations span a range from <0.006 eV [2] to 0.6 eV [4] for two similar Ni–Mn–Ga martensites. A Seeger-type phenomenological model of the resolved shear stress [12] has recently been proposed to explain the temperature dependence of the magnetic field $H$ and stress $\sigma$ induced transformation hysteresis in metamagnetic alloys of the Ni–Mn–X (X = In, Al, Sb) family [5,6] and Ni-rich NiTi alloys [7]. It was suggested that the temperature dependence of the $H$ and $\sigma$ hysteresis is due to an athermal $\sigma_\mu$ and a thermally activated component $\sigma_{TA}$. In classic dislocation theories, $\sigma_\mu$ is attributed to the long-range temperature-independent internal stresses, whereas $\sigma_{TA}$ (which is also referred to as effective stress $\sigma^*$) is ascribed to the short-range interactions with local obstacles [13]. The yield stress is given by:

$$\sigma_c(T) = \sigma_\mu + \sigma*(T) \quad (1)$$

The temperature dependences of $H$ and $\sigma$ hysteresis were fitted in Refs. 5–7 by an analytical function, Eq. (2), with 5 fitting parameters. Using additional data on the rate dependence of the transformation hysteresis, the activation energy of overcoming some kinetic barriers was estimated to be 0.7 eV [6,14] in the case of a Ni–Mn–In–Co alloy.

Lebedev and Kustov [15] reported that the temperature dependence of the macroscopic dislocation yield stress, $\sigma_c$, is proportional to that of the micro-yield stress, $\sigma_\varepsilon$. The micro-yield stress $\sigma_\varepsilon(T)$is derived from

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non-linear internal friction experiments as the stress, which produces a constant (arbitrary) level of reversible anelastic strain, $\varepsilon_{an}$. The proportionality is hence:

$$\sigma_c(T) = K\sigma_\varepsilon(T) \tag{2}$$

where K is a numerical factor depending on the selected $\varepsilon_{an}$ value.

This proportionality was found to hold for poly- and single crystals with different lattice types, impurity content, etc. In this paper we show that Eq. (2) holds also for twinned Ni–Fe–Ga–Co shape memory alloys where $\sigma_c$ is the macroscopic de-twinning stress and $\sigma_\varepsilon$ is the micro-yield stress. As Eq. (2) is at variance with the Seeger model [15] we will argue that it is the self-affine property of reversible motion of both 1D and 2D patterns (e.g. dislocations, twin boundaries) on different length scales that is the reason for the observed proportionality.

Single crystalline $[110]_A$ $Ni_{49}Fe_{18}Ga_{27}Co_6$ samples used in the present work were similar to those, for which a burst-like recovery of strain was studied [16]. Mechanical properties and microstructure of these crystals have been studied in detail [11,17–19]. Samples for macroscopic deformation and internal friction studies measured $4 \times 4 \times 9$ mm$^3$ and $1.5 \times 1.5 \times 21$ mm$^3$, respectively. The samples were annealed for 3 h at 1170 K and water quenched, resulting in the Curie temperature, $T_C$, of 375 K and martensitic transformation temperatures $M_s$, $M_f$, $A_s$ and $A_f$ of 265, 255, 271 and 280 K. The temperatures of phase transitions were determined from AC impedance measurements, performed in conventional 4 wire mode. After the heat treatment used the samples contained γ-phase particles and demonstrated B2 to 10 M–14 M transition [11,19].

The deformation was performed in compression in an Instron 1342 testing equipment with liquid He cryostat for temperatures between 4.2 and 240 K. The strain rate was $10^{-4}$ s$^{-1}$. The de-twinning stress was determined for a single sample at several temperatures during heating from 4.2 K. Deformation at each temperature incremented the overall strain of the sample by a small amount of approx. 0.1%. Internal friction (IF) was studied by means of a piezoelectric ultrasonic composite oscillator technique. The experimental arrangement is described in [20]. Longitudinal resonant oscillations of the sample where excited in a fundamental mode at a frequency around 90 kHz for temperatures between 17 and 300 K. The oscillator with the sample was cooled/heated in an Oxford closed loop cryostat. Temperature was registered by means of a Cernox sensor placed in close vicinity to the sample. The non-linear anelasticity emerging in the twinned martensitic phase will be attributed to the motion of intervariant and twin boundaries [21–24].

The internal friction, IF, is expressed as logarithmic decrement, δ, and determined as function of the elastic strain amplitude $\varepsilon_0$ (between $10^{-7}$ and $10^{-4}$) and temperature, $T$. The concomitant decrease of the effective Young's modulus (i.e. the 'Young's modulus defect') was derived from the resonant frequency of the sample (see Ref. 20 for details) simultaneously with $\delta(\varepsilon_0)$ measurements. This amplitude-dependent Young's modulus defect, ADMD, $(\Delta E/E)_h$, is determined by the ratio of the anelastic strain under maximum stress, $\varepsilon_{an}$, and $\varepsilon_0$ [25]:

$$\left(\frac{\Delta E}{E}\right)_h = \frac{\varepsilon_{an}}{\varepsilon_0} \tag{3}$$

The anelastic strain under maximum stress is then given by

$$\varepsilon_{an} = \varepsilon_0\left(\frac{\Delta E}{E}\right)_h \tag{4}$$

Fig. 1(a) displays the dependence of δ on elastic strain amplitude $\varepsilon_0$ for several temperatures between 239 and 17 K. Panel (b) shows, on an expanded scale, the data from Fig. 1(a) for temperatures below 120 K. During the measurements, $\varepsilon_0$ was first increased to the maximum value and then decreased in the reverse sequence. The difference between increasing and decreasing $\varepsilon_0$ remains small in all measurements.

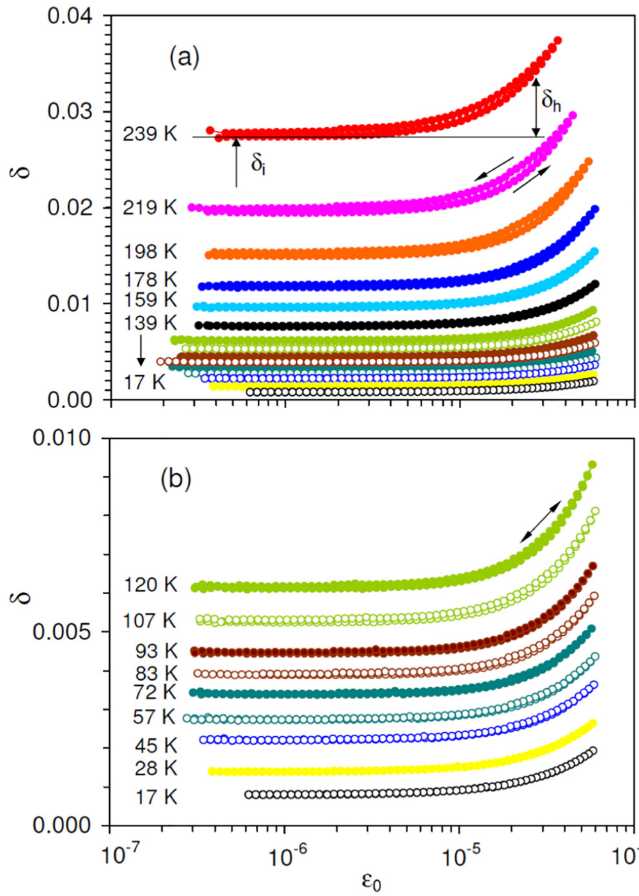


**Fig. 1.** Strain amplitude dependence of the logarithmic decrement δ of a sample of martensitic $Ni_{49}Fe_{18}Ga_{27}Co_6$ alloy registered at selected temperatures during cooling: a) overall view over the temperature range 239–17 K; b) low temperature range between 120 and 17 K on expanded scale. Each dependence consists of the points corresponding to increasing and decreasing strain amplitudes (increasing and decreasing strain amplitude scans are shown by arrows for curves at 219 and 120 K). The curve registered at 239 K is used to demonstrate decomposition of the total damping δ into the linear, $\delta_i$, and non-linear, $\delta_h(\varepsilon_0)$, components.

The results were fully reproducible in consecutive $\varepsilon_0$ scans, indicating that no irreversible changes are introduced by the mechanical oscillations. δ does not depend noticeably on $\varepsilon_0$ at strain amplitudes below ~$10^{-6}$. This low-amplitude linear IF is called $\delta_i$. The non-linear IF term, $\delta_h(\varepsilon_0)$, emerges at high strain amplitudes (e.g. Fig. 1(a)). The total damping is $\delta(\varepsilon_0) = \delta_i + \delta_h(\varepsilon_0)$ [26].

Fig. 2(a) shows the effect of temperature on $\delta_h(\varepsilon_0)$ when the background $\delta_i$ is subtracted from δ. $\delta_h(\varepsilon_0)$ is power law distributed with a temperature-independent stress exponent $n \approx 1.5$.

We can separate the $\varepsilon_0$ and $T$ dependences as:

$$\delta_h(T,\varepsilon_0) = A(T)\varepsilon_0^n \tag{5}$$

We now define a ratio $r$ between $\delta_h$ and $\left(\frac{\Delta E}{E}\right)_h$:

$$r = \frac{\delta_h}{(\Delta E/E)_h} \tag{6}$$

For example, in Fig. 2(b) we find $r$ ~ 1.3.

We now derive the micro-yield stress $\sigma_\varepsilon(T)$ from the non-linear IF data. We follow the same algorithm as was used in the case of dislocation (micro) plasticity [15], which requires:

find regimes in the $\delta_h(\varepsilon_0)$ curves where $\varepsilon_{an}$ is the same at different $T$,
determine the elastic strain amplitude $\varepsilon_0$ in these regimes,