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Transformation-induced plasticity as the origin of serrated flow in an NiTi shape memory alloy

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ABSTRACT

Transformation-induced plasticity was observed in an NiTi shape memory alloy. It was found that the austenite phase of a solutionized 50.5at.% Ni–Ti alloy shows a serrated stress–strain curve in a narrow range of 311–328 K, just below the temperature at which the deformation mechanism of the austenite phase changes from that of stress-induced martensitic transformation in the low temperature ranges to slip at high temperatures (329 K). Microstructure observation and thermal analysis have concluded the simultaneous occurrence of the two deformation mechanisms in the serrated flow. The interaction between the martensitic transformation and slip was investigated.

1. Introduction

The nickel titanium, NiTi or nitinol (Buehler and Cross, 1969), is a shape memory alloy (SMA) characterized by relatively large shape memory strain at around 7% in uniaxial tensile deformation in its polycrystals. In order to obtain a large shape recovery for the shape memory effect, slip is suppressed during deformation to provide the shape strain (pre-straining). However, resistance to slip may not be an inherent property of this alloy, since the austenite phase shows relatively high ductility when the phase is well homogenized (Scholl et al., 1968). Low temperature ductility is rare for B2 ordered intermetallic compounds (Poole and Hume-Rothery, 1955; Murray, 1987). When the homogeneous austenite phase undergoes martensitic transformation, the martensite is also ductile (Rozner and Wasilewski, 1966). Since the martensitic transformation preserves the slip plane of the austenite lattice as the basal plane of the B19' martensite (Knowles and Smith, 1981), the dislocation structures of the austenite phase are inherited to the martensite, and vice versa, as revealed by transmission electron microscope (Bataillard et al., 1998). As a result, it is reasonable to assume that both phases may have similar ductility, but not necessarily the same slip behavior.

It has been well demonstrated that the resistance to slip in the NiTi alloys can be improved by means of work hardening and precipitation hardening. The thermomechanical treatment of NiTi (Miyazaki et al., 1981a, 1982) involves making use of the two hardening effect when processing the material. Considerable effort has been made to develop thermomechanical treatment processes for this alloy (Todoroki and Tamura, 1987; Liu and McCormick, 1989). It has been clarified that cold working followed by annealing between 600 K and 800 K forms fine Ni_4Ti_3 precipitates which are responsible for the partial recovery of the dislocation substructure and the change in the matrix composition (Nishida et al., 1986; Lin and Wu, 1994). The size and density of the precipitates have been shown to influence the mechanical properties of NiTi significantly (Bataillard et al., 1998; Khalil-Allafi et al., 2002; Hamilton et al., 2004).

Thermomechanically processed alloys exhibit excellent shape memory effect and superelasticity. A typical stress–strain curve has a plateau without showing the strain-hardening during deformation by the variant change (detwinning) of

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martensite or the stress-induced transformation of the austenite phase. Further straining induces the slip of either the detwinned martensite or the stress-induced martensite (SIM). Thus, both the stress–strain curve of the martensite and that of the austenite phase are conventionally categorized into three stages, Stages I–III (Rozner and Wasilewski, 1966). Later works (Melton and Mercier, 1978, 1981; Miyazaki et al., 1981a, 1982; Perkins, 1984;Shaw and Kyriakides, 1995) have confirmed the deformation mechanism in each stage;

(i) Stage I occurs either by the detwinning of martensite (Liu et al., 1998; Sehitoglu et al., 2003) or by the stress-induced transformation in the austenite phases, depending on the temperature,
(ii) Stage II is the transient stage between Stages I and III. The great part is the elastic deformation of the detwinned martensite,
(iii) Stage III begins with the yielding due to the slip of either the detwinned martensite or the SIM.

The point is that Stages I and III need to be well separated by Stage II for a SMA to show large strain recovery. When slip is suppressed sufficiently in Stage I, the majority of the pre-strain exhibits a shape memory effect.

Even when the shape memory effect appears perfect, however, it is inevitably accompanied by a small amount of plastic strain. This is pronounced in the first few cycles of the shape memory effect: the training process, or after several repetitions, referred as the cycling process (Melton and Mercier, 1979; Perkins, 1984; Miyazaki et al., 1986; Xie et al., 1998; Gall and Maier, 2002). In order to describe this kind of plastic strain, called microplasticity, Bilby proposed the concept of transformation dislocation (1953). The dislocation is non-glissile in type and is generated by the incoherency of either the martensite/austenite boundary or the martensitic twin boundary (Kajiwara and Kikuchi, 1982; Christian, 1994). The plastic strain originating from dislocations of this type is small compared with the slip responsible for the crystal plasticity in regular metals. The latter plasticity includes the multiplication of a large number of dislocations and the collective motion of the long-range path as described by the Bailey–Hirsch equation (Seeger and Kronmuller, 1962).

Microplasticity also occurs during the growth of thermal martensite, giving rise to the hysteresis of transformation temperature. Lovey et al. (2004) presented a model of the interaction between a few isolated dislocations and a single martensite plate responsible for the increase in the width of hysteresis. It may be that the stress hysteresis in superelasticity has the same origin. It is also known that the transient heat transfer of latent heat is the other major origin of the increase in hysteresis. This has been clarified both experimentally (Shaw and Kyriakides, 1995) and also in modeling studies (Christ and Reese, 2009; Morin et al., 2011).

In recent years, a slip similar to Stage III was observed in Stage I by the compression of submicron-sized NiTi pillars (Frick et al., 2007; Frick et al., 2008; Norfleet et al., 2009; Simon et al., 2010; Manjeri et al., 2010). This experimental technique makes it possible to measure the strength of a submicron-sized specimen machined from bulk material by means of the focused ion beam. Frick et al. (2007, 2008) determined that the loss of superelasticity resulted from slip in the pillar. Nordfleet et al. (2009) observed that both slip and SIM occur simultaneously, and referred to the phenomenon as transformation-induced plasticity (TRIP), which is observed in steels (Grässel et al., 2000).

The research reported herein was carried out to investigate whether slip could result from the loss of strength after micro-machining. Thermomechanical treatment is done in the bulk state prior to machining. Since the diameter of the pillar decreased to somewhere in the order of a few nanometers, a size comparable to or smaller than the average spacing of the Ni_4Ti_3 precipitates (Nishida et al., 1986; Khalil-Allafi et al., 2002; Gall and Maier, 2002; Hamilton et al., 2004), the precipitates should not function as obstacle of slip (*e.g.* Hull and Bacon, 1984), and slip should occur readily in the pillar. The results of an investigation into the grain size dependence of the yield stress in a nano grained NiTi alloy (Delville et al., 2011) support this hypothesis.

In the last two decades, noticeable advances have been made in understanding the continuum mechanics of SMA. An early study of the phenomenological modeling by Tanaka (1986) should be acknowledged. An excellent overview was given in the paper of Lagoudas and Entchev (2004), who classified the literature into two groups: micromechanical-based models and phenomenological models. Readers may refer to recently published articles for up-to-date knowledge of finite deformation analysis (Reese and Christ, 2008), polycrystal plasticity (Peng et al., 2008), the micromechanics-based approach (Jemal et al., 2009), a path-dependent analysis on the basis of thermomechanical modeling (Sedlák et al., 2012), and the thermodynamic constitutive model (Song et al., 2013). It should be noted that several authors have dealt with the evolution of plastic strain, which is closely related to the present subject: the three dimensional thermomechanical constitutive model (Lagoudas and Entchev, 2004): a kinematic and isotropic hardening model (Paiva et al., 2005): the finite element simulation of three dimensional aggregate of cubic grains (Wang et al., 2008; Manchiraju and Anderson, 2010): the simulation of ratcheting behavior (Kan and Kang, 2010): viscoelastic models of crystal plasticity (Hartl et al., 2010; Yu et al., 2012).

It has already been recognized that the plastic strain adopted in these modeling studies differs from the crystal plasticity in regular metals (Lagoudas and Entchev, 2004). This crystal plasticity was mostly due to the transformation dislocation generated in training or cycling process, as mentioned above. The yield stress could be as low as the flow stress of the variant deformation (detwinning) of martensite. However, it is difficult to determine the yield stress experimentally because of the difficulty in dealing with the interaction between the martensitic transformation and slip. It should be noted that the observation of slip in monotonous tensile testing has been limited to a few papers (Miyazaki et al., 1982; Kato et al., 1999; Sehitoglu et al., 2001).