Strain Dependence of Pseudoelastic Hysteresis of NiTi

YINONG LIU, I. HOUVER, H. XIANG, L. BATAILLARD, and S. MIYAZAKI

This work investigated the transformation-strain dependence of the stress hysteresis of pseudoelasticity associated with the stress-induced martensitic transformation in binary NiTi alloys. The strain dependence was studied with respect to the deformation mode during the stress-induced martensitic transformation, which was either localized or homogeneous. It was observed that the apparent stress hysteresis of pseudoelasticity was independent of the transformation strain within the macroscopic deformation range, for the specimens deformed in a localized manner. For specimens macroscopically deformed uniformly, the stress hysteresis of pseudoelasticity increased continuously with increasing strain from the beginning of the stress-induced martensitic transformation. The transformation-strain independence of the stress hysteresis for localized deformation is ascribed to be an artificial phenomenon, whereas the transformation-strain dependence of the hysteresis for uniform deformation is believed to be intrinsic to the process of stress-induced martensitic transformation in polycrystalline materials. This intrinsic behavior is attributed to the polycrystallinity of the materials.

I. INTRODUCTION

THE stress hysteresis of pseudoelasticity associated with thermoelastic martensitic transformations in shape-memory alloys has been reported in the literature to be either dependent or independent of the transformation strain under various conditions. In the former case, the magnitude of the stress hysteresis is found to increase with the strain amplitude of a pseudoelastic loop.[1,2,3] In the latter case, the hysteresis is observed to be independent of the strain of the pseudoelastic loop within the strain limit of a stress-induced martensitic transformation.[4,5] The transformation-strain independence of pseudoelastic hysteresis is often observed in tensile testing using specimens of large length-to-diameter aspect ratios for binary NiTi alloys.[4,5] In shear deformation of similar alloys, however, a transformation strain–dependent pseudoelastic hysteresis has been observed.[2] This uncertainty has caused some inconsistency in the effort of modeling the thermomechanical behavior of shape-memory alloys and the understanding of the thermodynamics of thermoelastic martensitic transformations.

It has been well established that the process of a thermoelastic martensitic transformation, either thermally induced or stress induced, is accompanied by an internal elastic energy and an irreversible energy.[6,7] The irreversible energy is believed to be responsible for the hysteretic behavior of the transformation. The dissipation of the irreversible energy during the course of a thermoelastic martensitic transformation is due to the need to overcome the resistance from an imperfect matrix to the diffusionless, displacive movement of a transformation-phase boundary. It is generally regarded that this resistance, originating from structural defects, is determined by the metallurgical conditions of the matrix and, thus, the irreversible energy is a material constant. Consequently, the transformation hysteresis, either the temperature hysteresis of a thermal transformation cycle or the stress hysteresis of a mechanical transformation cycle, is expected to be a constant, ignoring possible modifications to the microstructure during one individual transformation cycle. Based on this understanding, irreversible energy has been estimated from experimental measurements of transformation hysteresis, and the estimated values have been used in thermodynamic analysis.[8,9,10] Therefore, it is of direct interest and primary importance to clarify the uncertainty about the transformation-strain dependence of the stress hysteresis of pseudoelasticity associated with thermoelastic martensitic transformations. This work was carried out to investigate the effect of the deformation mode, which was either localized or uniform, on the strain dependence of pseudoelastic hysteresis of polycrystalline near-equiatomic NiTi alloys.

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II. EXPERIMENTAL PROCEDURE

Two commercial binary NiTi alloys were used, both with a nominal composition of 50.2 at. pct Ni. Alloy I was received in wire form, 1.5 mm in diameter, and alloy II was in plate form, 1.3 mm in thickness. The wire and plate specimens were fabricated from these two alloys, respectively. Wire specimens were prepared by cold rolling alloy I in multiple steps to 0.87×0.87 mm^2 in cross section. Plate specimens were fabricated from alloy II, after cold rolling, by grinding into a "dog-bone" shape with a gage length of ~20 mm and a gage cross section of 1.0×5.3 mm^2. The gage length of the plate specimens was in the rolling direction. The specimens fabricated were annealed at different temperatures between 673 K and 723 K to obtain different pseudoelastic characteristics. The transformation behavior of the specimens was measured by differential scanning calorimetry (DSC) using a PERKIN-ELMER* Differential Scanning Calorimeter 4.

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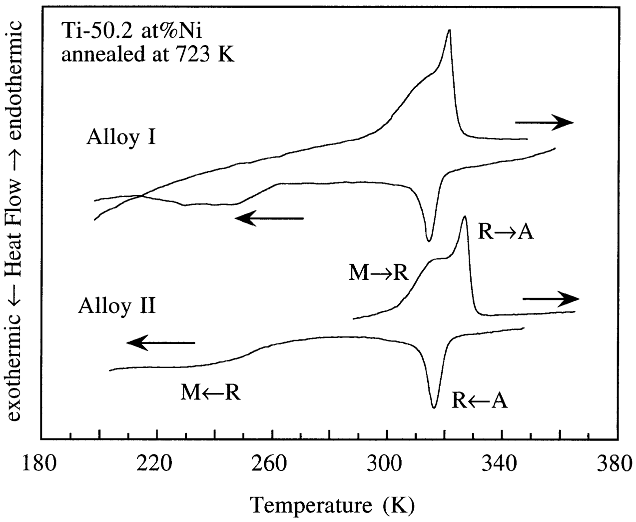

Fig. 1—DSC measurement of the transformation behavior of the two NiTi alloys. The two specimens exhibited identical transformation behavior with a two-step A→R→M transformation on cooling and a two-step M→R→A transformation on heating.

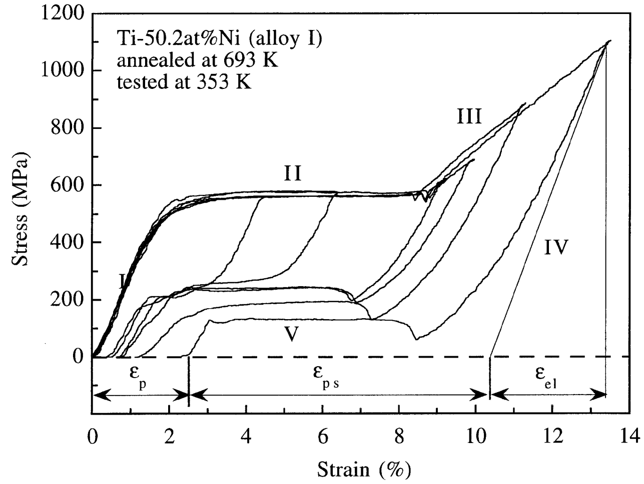

Fig. 2—Tensile pseudoelastic behavior of the wire specimens. The specimens exhibited typical localized deformation behavior. The deformation stages designated by roman numerals are identified in the text.

Mechanical testing was carried out using an Instron 4301 testing machine. Pressure grips were used for the plate specimens. For wire specimens, two different types of grips were used: pressure grips, which gripped directly onto the bare wire specimen, and compression-free "hook" grips, which enabled only tensile loading to be applied to the specimen. The specimens used with the hook grips were fitted with special ends for gripping. The Instron machine was equipped with a liquid bath for temperature control. The use of the liquid bath prevented the measurement of local deformation using an extensometer; thus, the displacement of the crosshead was recorded for deformation measurement. The crosshead displacement was calibrated relative to a local strain measurement using an Instron 2630-100 extensometer on a steel plate specimen. The difference between the two measurements was found to be negligible at low deformation levels, typically 0.1 to 0.2 pct error at a strain level of 7 pct and a stress level of ~200 MPa. The error increased with increasing force, reaching 3 pct at ~5000 N. This is attributed to the fact that a portion of the steel specimen that was initially clamped between the pressure grips was stretched out of the grips at high force levels, which effectively enlarged the gage length of the deformation. This resulted in the global measurement being greater than the true (local) strain within the gage length.

III. RESULTS

A. *Transformation Behavior*

Figure 1 shows DSC measurements of the transformation behavior of the two alloys after an identical heat treatment at 723 K for 3.6 ks. The transformation behavior of the two specimens was practically identical. Both specimens exhibited a two-stage transformation on cooling, corresponding to an austenite–to–*R* phase ($A \rightarrow R$) transformation at the higher temperature and an *R* phase–to–martensite ($R \rightarrow M$) transformation at the lower temperature. The specimens also exhibited a two-stage transformation on heating, corresponding to a martensite–to–*R* phase ($M \rightarrow R$) and an *R* phase–to–austenite ($R \rightarrow A$) transformation, sequentially. The two transformations on heating were much less separated as compared to their forward transformations on cooling. The characteristic temperatures for the transformations for alloy I, as determined at the maximum heat flow, are measured to be $T_{A \rightarrow R} = 316$ K, $T_{R \rightarrow M} = 241$ K, $T_{M \rightarrow R} = 316$ K, and $T_{R \rightarrow A} = 327$ K.

It is seen that the $R \rightarrow M$ transformation occurred over a very wide temperature interval with a very low heat-flow intensity for the specimens annealed at 713 K. For specimens annealed at lower temperatures, no $R \rightarrow M$ transformation was observed within the temperature range from 180 to 380 K for the DSC measurement. This observation is in agreement with previous studies.[8,11]

B. *Tensile Deformation*

Figure 2 shows the pseudoelastic stress-strain curves of wire specimens deformed to different strain levels. The specimens were annealed at 693 K for 3.6 ks after cold working. The specimens shown in this figure were bare wire specimens gripped using pressure grips. The pseudoelastic stress-strain curves of these specimens may be divided into several stages, according to the apparent difference in deformation mechanisms. Stage I corresponds to the initial loading section, which is commonly recognized as the elastic deformation of the austenite.[12,13,14] Stage II occurred over a stress plateau, which is associated with a stress-induced martensitic transformation. The end of the stress plateau is often regarded as the end of the stress-induced martensitic transformation. Stage III proceeds between the end of the stress plateau and the onset of a second apparent yielding. It has been suggested that it is not a stage of pure elastic deformation of the stress-induced martensite,[15,16] as evidenced by the deviation of the unloading curve from the loading curve. The exact mechanism of deformation in this stage is yet to be fully understood. The unloading section, denoted as stage IV, is generally regarded as the elastic unloading of the

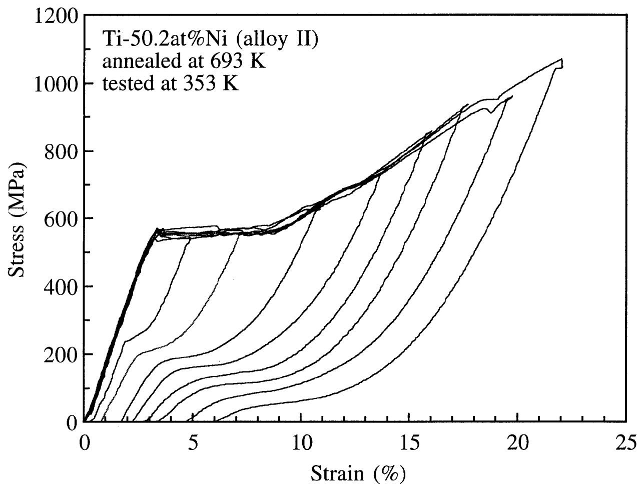

Fig. 3—Tensile pseudoelastic behavior of the plate specimens. The specimen exhibited localized deformation behavior during loading and uniform deformation behavior during unloading.

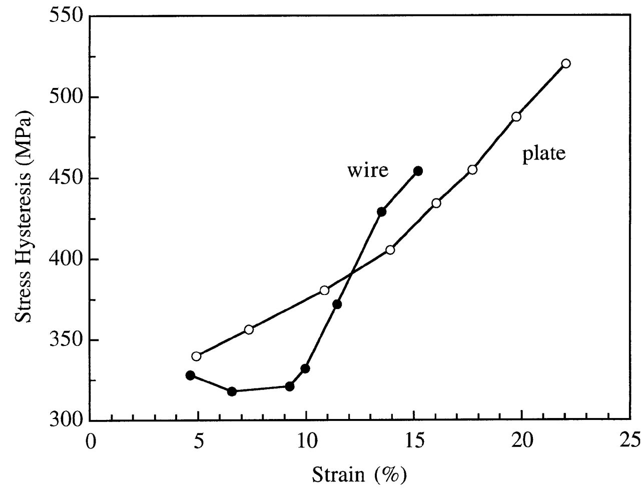

Fig. 4—Effect of deformation mode on stress hysteresis of pseudoelasticity, showing different strain dependences of the hysteresis for the two specimen geometries.

stress-induced martensite prior to the reverse transformation.[12,17] The unloading curve is frequently observed to be nonlinear, as evident in the figure. This imposes a challenge to the elastic hypothesis. The reverse transformation occurred over another stress plateau during unloading, denoted as stage V. The stress difference between the upper and the lower stress plateaus is the mechanical (stress) hysteresis of the pseudoelasticity.

As shown in Figure 2, the wire specimens exhibited a typical Lüders-type deformation behavior for both the forward and the reverse transformations during loading and unloading, respectively. The stress plateau associated with the forward transformation terminated with a stress drop at the end of the stress plateau for the forward transformation, clearly marking the transition of deformation from stage II to stage III. The Lüders-type deformation behavior upon unloading was characterized by a pronounced inverse stress peak at the beginning of stage V for specimens deformed to beyond the stress plateau during the forward deformation. For specimens in which the forward deformation was terminated prior to the end of the stress plateau (stage II), no inverse stress peak was observed upon unloading.

It is to be mentioned that a stress peak was absent at the beginning of the upper stress plateau for the forward stress–induced martensitic transformation, although it is commonly observed in tensile deformation of specimens of large aspect ratios. This is due to the direct compression gripping of the specimens adopted in this experiment. Under this gripping condition, some stress-induced martensite had already been formed at the ends of the gage section of the specimen prior to the tensile deformation.

Figure 3 shows the pseudoelastic stress-strain curves of the plate specimens. The specimens were annealed at the same temperature as the wire specimens shown in Figure 2. It is seen that the plate specimens also exhibited a Lüders-type deformation during the forward stress–induced martensitic transformation. The transition from stage II to Stage III, however, was less obvious compared to the wire specimens. The stress plateau for the forward transformation was characterized by a stress peak at the beginning. This is due to the fact that the martensite that had been induced by the compression of the pressure grips was outside the gage section of the specimen. Therefore, the stress-induced martensite in the grips did not affect the stress-induced martensitic transformation of the specimen within the gage section. The reverse transformation during unloading occurred in a uniform manner, as characterized by the smooth, continuously decreasing stress and the absence of the inverse stress peak.

Figure 4 shows the measurements of the stress hysteresis of pseudoelasticity, as a function of the strain, for both the wire and the plate specimens. The stress hysteresis of the plate specimens increased continuously with increasing strain. The stress hysteresis of the wire specimens remained approximately constant with strain at low strain levels within the limit of the stress plateau and increased with increasing strain at higher strain levels.

Measurements of strain recoveries are shown in Figure 5. The total recovered strain consisted of two components, the elastic recovery (ε_{el}) and the pseudoelastic recovery (ε_{ps}),

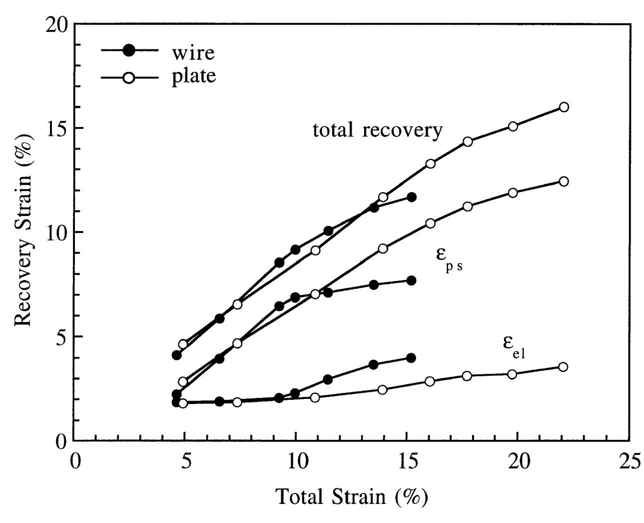

Fig. 5—Effect of total deformation on shape recovery for the two specimen geometries.

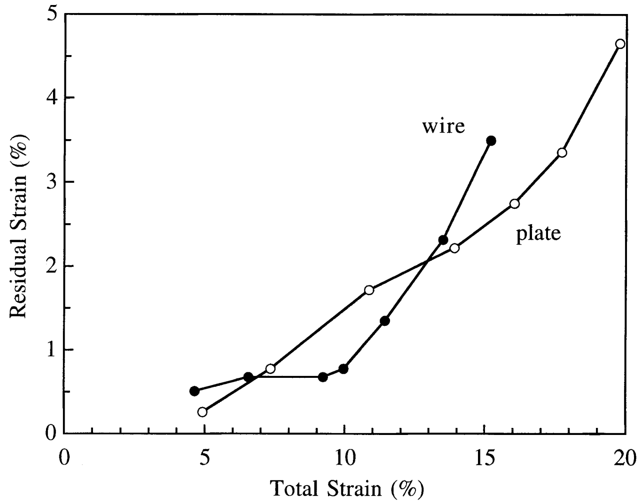

Fig. 6—Effect of deformation on residual strain of pseudoelasticity.

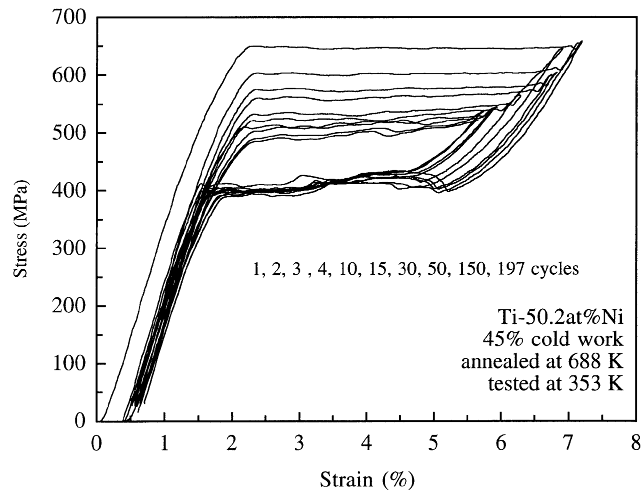

Fig. 7—Pseudoelastic cycling of a specimen annealed at 688 K. The specimen exhibited Lüders-like deformation behavior during the stress-induced martensitic transformation throughout the cycling.

as indicated in Figure 2. The elastic recovery was estimated using a Young's modulus of 30 GPa[18] for the martensite.*

*It is known that the apparent modulus of elasticity of the martensite in NiTi, as measured in mechanical testing, varies with a number of factors, indicating that the deformation mechanisms other than elasticity contribute to the apparently linear deformation of the martensite. Under this condition, the maximum value measured is regarded as the value closest to the true value of the modulus of elasticity, provided that the measurement is reliable.The value of 30 GPa is a conservative one. Using a conservative value ensures that the elastic strain is not underestimated and the pseudoelastic recovery is not overestimated.

The elastic recovery remained constant at a low level at strain levels up to 10 pct, due to the constant stress at the stress plateau of the stress-induced martensitic transformation. The elastic recovery increased with strain at higher strain levels due to the increased stress in stage III. The pseudoelastic recovery of the plate specimens continued to increase with deformation after the stress plateau. For the wire specimens, the increase of ε_{ps} at strain levels above 10 pct was more moderate than that of the plate specimens.

Measurements of the residual strain are shown in Figure 6. The residual strain of the plate specimens was found to increase continuously with total strain. The residual strain of the wire specimens, on the other hand, remained approximately constant at strain levels below 10 pct and increased rapidly at higher strain levels, in a similar manner to the strain dependence of the stress hysteresis.

C. *Pseudoelastic Cycling*

The results shown previously indicated that the pseudoelastic hysteresis was independent of transformation strain for the wire specimens, which deformed in a localized manner both during the forward transformation and the reverse transformation, whereas the hysteresis was dependent on transformation strain for the plate specimens, which deformed in a uniform manner during the reverse transformation. To eliminate the uncertainty that may arise with the use of two different materials with two possibly different textures, another two wire specimens were prepared, which were subjected to pseudoelastic cycling to achieve a different deformation behavior. In this test, the hook grips were used so that the risk of compression on the specimen was avoided during pseudoelastic cycling.

Figure 7 shows the evolution of the pseudoelastic behavior of a wire specimen during pseudoelastic cycling up to 200 cycles. The curves shown in the figure are selected numbers of cycles, as indicated. The specimen was annealed at 688 K for 1.8 ks after cold rolling. The cycling was conducted at 353 K in water, between zero stress and the strain limit of the stress plateau for the forward transformation. The specimen exhibited a typical localized deformation behavior during both the forward and reverse transformations, with a well-defined stress plateau throughout the cycling. During cycling, the critical stress for the forward transformation decreased progressively, while the critical stress for the reverse transformation remained approximately unchanged, leading to a reduced stress hysteresis. The stress hysteresis of the initial pseudoelastic cycle was 250 MPa, whereas, after 200 cycles, the hysteresis was measured to be 80 MPa. Accompanying the decrease of the critical stress for the forward transformation, the magnitude of strain of the stress plateau also decreased, from 7 pct in the first pseudoelastic cycle to 5.1 pct after 200 cycles. The specimen showed a nearly perfect pseudoelasticity, with a small accumulated residual strain of $<$0.5 pct after 200 cycles. After the cycling, the stress-strain behavior of the pseudoelasticity became stabilized and highly repeatable.

Pseudoelastic cycles to different strain levels were performed for the measurement of the strain dependence of stress hysteresis after the pseudoelastic cycling, as shown in Figure 8. The pseudoelastic cycles were performed sequentially from the cycle of minimum strain to that of the highest strain. It was observed that complete shape recovery was achieved for deformation up to 8 pct, well beyond the strain limit of the stress plateau, which terminated at 5.4 pct of strain after the cycling.

For comparison, the second wire specimen was also cycled in pseudoelasticity at 353 K for 30 cycles, as shown in Figure 9. This specimen was annealed at 713 K for 1.8 ks after cold rolling. An increased annealing temperature enabled the specimen to change its deformation behavior from a Lüders-like, localized manner to a uniform manner

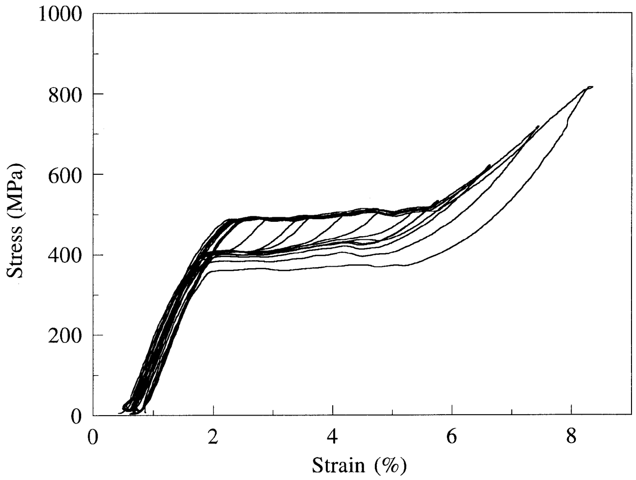

Fig. 8—Pseudoelastic loops for the measurement of hysteresis after 200 cycles. The specimen showed a nearly complete shape recovery up to 8 pct of deformation.

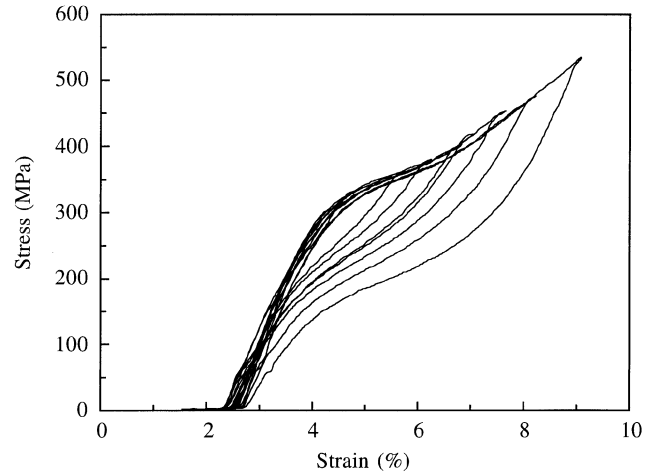

Fig. 10—Pseudoelastic loops for the measurement of hysteresis after 30 cycles.

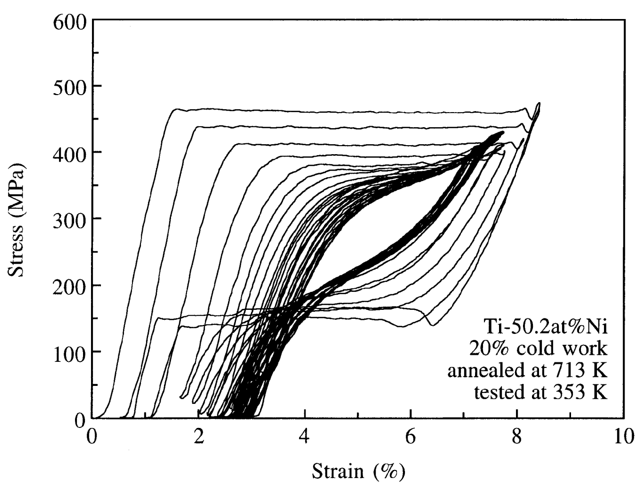

Fig. 9—Pseudoelastic cycling of a specimen annealed at 713 K. The Lüders-like deformation of the specimen evolved gradually into a uniform deformation during the cycling.

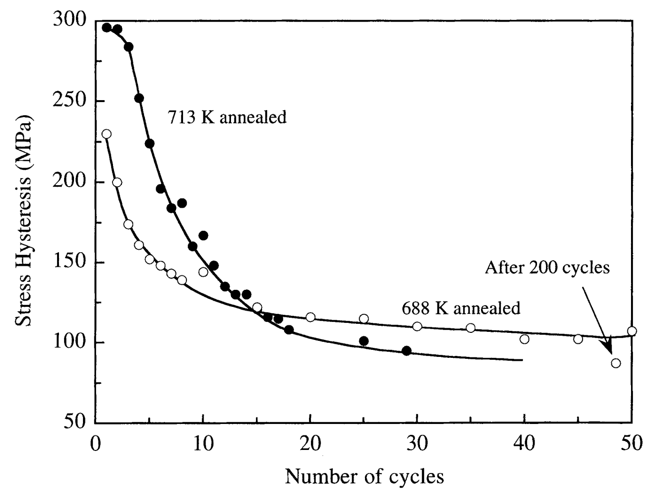

Fig. 11—Evolution of stress hysteresis during pseudoelastic cycling. The hysteresis decreased progressively to much lower values after the cycling.

during pseudoelastic cycling, as evident in the figure. The stress hysteresis of the first pseudoelastic cycle was measured to be 305 MPa. This specimen also showed a significant increase in the residual strain accumulated during the cycling. A residual strain of ~3 pct was measured after 30 cycles. After 30 cycles, the stress-strain curve became stabilized and repeatable, and pseudoelastic cycles deformed to different levels of strain were performed for the measurement of the strain dependence of stress hysteresis, as shown in Figure 10. The measurement was performed, progressively, from cycles of low strain levels to those of high strain levels.

Figure 11 shows the evolution of the stress hysteresis of pseudoelasticity during pseudoelastic cycling for the two specimens. The stress hysteresis for the specimen annealed at 688 K was easily determined to be between the upper and lower stress plateaus. For the specimen annealed at 713 K, the hysteresis was determined to be the maximum difference of stress between the forward transformation and the reverse transformation at any given strain, *i.e.*, no compensation for elastic deformation at different stress levels was adopted. The measurements show that the stress hystereses decreased rapidly during initial cycles and saturated at much-lower levels after cycling. The stress hysteresis of the specimen annealed at 713 K was initially greater than that of the specimen annealed at 688 K, but was smaller after a number of cycles.

Figure 12 shows the measurements of the stress hysteresis of pseudoelasticity as a function of the strain span of the pseudoelastic loops for the two specimens, as determined from the stress-strain curves shown in Figures 8 and 10. For the specimen annealed at 688 K, the stress hysteresis remained approximately constant with deformation strain up to 5.4 pct, the strain limit of the upper stress plateau, and increased with strain at higher deformation levels. In contrast, the stress hysteresis of the specimen annealed at 713 K increased continuously with strain through the entire range, exhibiting a near-linear dependence on the strain.

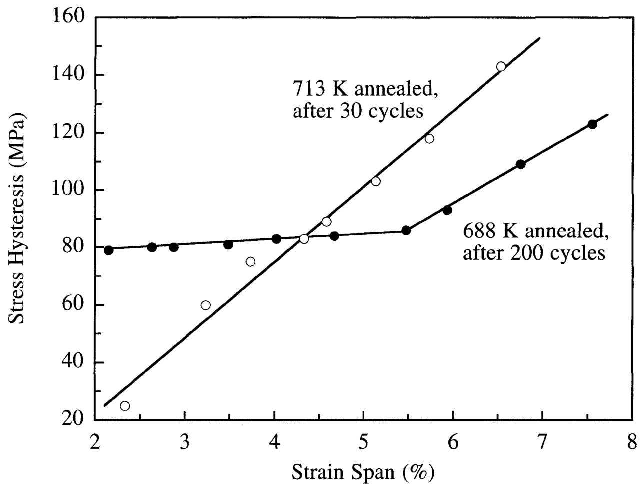

Fig. 12—Dependence of stress hysteresis on the strain span of pseudoelastic loops. The hysteresis is independent of the strain with localized deformation and increases with strain with uniform deformation.

IV. DISCUSSIONS

A. *Deformation Behavior of Stress-Induced Martensitic Transformation*

It is generally perceived that the stress-induced martensitic transformation in a given material occurs at a constant stress under isothermal conditions. This largely stems from the experimental observation of a stress plateau that is recognized to be associated with the stress-induced martensitic transformation, as evident in Figures 2 and 7. This concept has been formulated in thermodynamics in the free-energy balance consisting of chemical and nonchemical contributions.[6,7] The experimental observation of such a behavior has often been interpreted as an indication that the irreversible component of the nonchemical free-energy change, which is responsible for the hysteretic behavior of the transformation, is a constant during the process of a transformation.

It has, however, also been observed that a stress-induced martensitic transformation may proceed in a uniform manner, as is evident in Figure 10, under a variety of conditions, including deformation in shear,[2] in compression,[19] and after pseudoelastic cycling.[20] The implication of this observation is in direct contradiction to the concept of constant driving force and constant irreversible energy for a stress-induced martensitic transformation.

These experimental observations demonstrate that a stress-induced martensitic transformation may proceed, macroscopically, in either a localized manner or a uniform manner, although that, on a microscopic scale, a stress-induced martensitic transformation can only be localized, as in the case of any first-order phase transformation. In the case of (macroscopic) uniform deformation during stress-induced martensitic transformation, the hypothesis of constant irreversible energy and, thus, constant hysteresis is invalidated.

Another aspect concerning the deformation behavior of stress-induced martensitic transformation is that a nearly complete pseudoelastic recovery may be achieved with deformation well beyond the end of the stress plateau, as is evident with the specimen shown in Figure 8. Whereas it is generally recognized that the stress plateau is associated with a stress-induced martensitic transformation, the continued increase of pseudoelastic recovery with deformation in stage III suggests that more oriented martensite was produced. Similarly, for the specimen deformed uniformly during the stress-induced martensitic transformation, the pseudoelastic recovery increased with strain to beyond the point of inflection on the stress-strain curve, as shown in Figure 10. These observations are in agreement with previous studies.[4,16] It is also seen in Figure 7 that the strain span of the stress plateau decreased with pseudoelastic cycling, from 4.8 to 3.3 pct after 200 cycles. This also suggests that the end of the stress plateau does not correspond to the full realization of the transformation strain, which is fundamentally determined by the crystallography of the transformation, at least for the specimen in the state after the cycling. All these observations indicate that more preferentially oriented martensite was produced by the external stress in stage III.

The production of oriented martensite in stage III may be the result of either of two possible mechanisms: (1) further stress-induced martensitic transformation from some residual austenite, or (2) reorientation of some martensite variants induced by the external stress during stage II. The occurrence of the first mechanism indicates that the end point of the stress plateau is not the end of the stress-induced martensitic transformation, whereas the operation of the second mechanism implies that martensite variants oriented in directions other than the preferential orientation are produced by the uniaxial external stress during the stress-induced transformation. In either case, a further explanation is required.

B. *Transformation-Strain Dependence of Stress Hysteresis of Pseudoelasticity*

The experimental observations shown in Figures 2, 3, 7, and 9 may be summarized as follows. For the specimens deformed in a Lüders-like manner, both during the forward stress–induced martensitic transformation and the stress-restrained reverse transformation of the martensite, the stress hysteresis of the pseudoelasticity was independent of the transformation strain within the range of localization of deformation. In stage III, when the deformation became uniform, the hysteresis increased with strain. For the specimen exhibiting a uniform deformation both during the forward and the reverse transformation, the stress hysteresis increased with strain in the entire range of deformation. For the specimens deformed in a localized manner during the forward transformation and a in a uniform manner during the reverse transformation, the pseudoelastic hysteresis increased with strain, exhibiting a similar dependence on strain to that of the specimen deformed uniformly. The strain dependence of the stress hysteresis may change even in the same specimen, if its deformation behavior is changed from being localized to being uniform.

The strain independence of the stress hysteresis of pseudoelasticity, observed in the specimen which exhibited a Lüders-type behavior, is believed to be artificial. In a Lüders-like deformation process, the global strain increases with the propagation of deformation bands,[12,13,21] which are, in this case, the bands of stress-induced martensite. It has been established that the deformation within a localized deformation band corresponds to the strain at the end of the stress plateau, whereas that in the regions outside the localized deformation bands corresponds to the strain at the onset of the stress plateau.[21] At different stages of deformation within the range of the stress plateau, the total length of the bands of stress-induced martensite is different, but the microstructural and mechanical conditions inside the regions of nontransformed austenite and the regions of stress-induced martensite remain the same, respectively. That means that the conditions at the deformation interfaces, which are the boundaries of localized transformation bands, are identical throughout the process of deformation within the strain limit of the stress plateau. It is for this reason that the transformation deformation proceeds over a stress plateau. On reversion, the same interface moves back, regardless of the position of the interface within the gage length of the specimen. This leads to a stress hysteresis that appears independent of the magnitude of the global strain. This hysteresis is the hysteresis for the movement of an interface between a region of stress-induced martensite (end of the stress plateau) and a region of austenite (beginning of the stress plateau). In this context, the measurement of the stress hysteresis at strain levels lower than the strain limit of the stress plateau is false with respect to the apparent global strain of deformation.

When a specimen is macroscopically deformed uniformly during the stress-induced martensitic transformation, the critical stress increased with increasing strain for the forward transformation and decreased with strain for the reverse transformation, as shown in Figure 10. In this case, the recorded external stress is the driving force required to activate many microscopic, internal transformation interfaces in the matrix of the specimen. Although each of these transformation interfaces is still an interface between a martensite and an austenite, the mechanical conditions in the vicinity of an interface, *e.g.*, the stress field and geometry, are changing continuously during the process of the stress-induced transformation, with the growth in size and the increase in number of martensite plates. Therefore, the external stress recorded in this case reflected the true driving force for the transformation corresponding to that particular stage of transformation. In this context, the stress hysteresis measured in this case is regarded as the true hysteresis with respect to transformation strain. This observation that the stress hysteresis increased with the strain is consistent with previous measurements of NiTi alloys in homogeneous deformation cycles, both in pseudoelasticity and ferroelasticity.[2]

It is, however, necessary to point out that such a stress hysteresis determined from a stress-strain curve of a polycrystalline specimen is only the apparent, macroscopic hysteresis of pseudoelasticity. There has been no experimental evidence proving that the upper stress and the lower stress at a given strain level correspond, respectively, to the forward and the reverse movement of an identical interface. Therefore, such a hysteresis cannot be interpreted as a measure of the internal resistance to the movement of a transformation-phase boundary. It requires careful consideration to relate the measurement of an apparent macroscopic hysteresis to the internal resistance to a transformation phase–boundary movement, *e.g.*, for thermodynamic analysis.

C. *Mechanisms of the Strain Dependence of Pseudoelastic Hysteresis*

While it is recognized, in the previous discussion, that the strain dependence of pseudoelastic hysteresis observed in the case of uniform deformation during stress-induced martensitic transformation reflects the true relationship between the global strain and the macroscopic hysteresis for a polycrystalline matrix, the mechanisms for this relationship are yet to be established.

In a thermal process of the martensitic transformation, the only irreversible resistance that causes the thermal hysteresis is the frictional force to the transformation phase–boundary movement. In a stress-induced martensitic transformation in a polycrystalline matrix, however, at least one additional contribution in addition to the internal frictional resistance to phase–boundary movement is effective in causing a hysteresis: the mechanical resistance to a shape change. The existence of such a mechanical resistance to a shape change may be envisaged by considering a stress-induced martensitic transformation at the M_s temperature for a polycrystalline material. Should this mechanical resistance be zero, a stress-induced martensitic transformation would be expected to take place at virtually zero stress at the M_s temperature, where the thermal driving force for the transformation is high enough to overcome the frictional resistance to phase-boundary movement. In experimental practice, however, a minimum stress is always measured for a stress-induced martensitic transformation.[22]

The mechanical resistance against the movement of the interface between a martensite and a parent phase will be zero at the M_s temperature when the transformation is induced thermally. In this case, several martensite variants with different orientations will be formed to self-accommodate each other, to minimize their elastic strain energy. However, in the case of a stress-induced martensitic transformation in a polycrystalline matrix, only a preferentially oriented martensite variant in each grain will be formed under the assistance of an applied stress, so that the elastic strain will not be reduced by a self-accommodation mechanism, causing the mechanical resistance to exist even at the M_s temperature.

It has been suggested that it is impossible to achieve a fully oriented martensite, either by stress-induced martensitic transformation or by martensite reorientation, in a polycrystalline matrix without plastic deformation.[16] This is because of the orientation differences among the preferentially oriented martensite variants in neighboring austenite grains, which are of different crystallographic orientations. This orientation mismatch among the preferential variants in neighboring austenite grains requires a mechanism of deformation to operate at grain boundaries, to compensate for the discontinuity of internal microscopic deformation of the preferentially oriented variants and to maintain the integrity of the matrix. This mechanism of internal deformation may be either the formation of some martensite variants deliberately oriented in directions other than the preferential orientation or plastic deformation at junctions of some preferentially oriented variants of martensite. In the former case, the conversion of the "misoriented" variants to the preferential orientation requires internal plastic deformation to replace their role as the orientation mismatch coordination mechanism. In fact, it has been found in transmission electron microscopic observations that internal plastic deformation is induced at junctions of martensite variants by martensite reorientation deformation within the strain limit of the stress plateau.[23]

Clearly, this coordinating mechanism is only required in polycrystalline matrices. In a single-crystal specimen, no such coordination is required, because the lattice distortion associated with the transformation can be released to the free surface. In a polycrystalline matrix, the demand for such internal plastic deformation increases with increasing levels of orientation or formation of oriented martensite, in the case of stress-induced martensitic transformation. This is consistent with the experimental observation that the stress increased continuously with strain during the process of stress-induced martensitic transformation.

V. CONCLUSIONS

1. The experimental results of this study provide further evidence that the end of the stress plateau does not correspond to the end of the production of transformation strain during the process of a stress-induced martensitic transformation. The deformation mechanism after the stress plateau is recognized to be a mixture of three processes: the elastic deformation, plastic deformation, and further production of oriented martensite. The production of oriented martensite in this stage may be either of two possible mechanisms: the reorientation of some misoriented martensite variants formed during the deformation over the stress plateau, or the stress-induced martensitic transformation of some residual austenite. The present experiment does not provide evidence to differentiate these two mechanisms.
2. The stress hysteresis of pseudoelasticity is dependent on the magnitude of pseudoelastic deformation in polycrystalline materials. Such a dependence is observed when the deformation associated with the stress-induced transformation proceeds in a macroscopically uniform manner. However, the difference between the critical stresses for the forward and the reverse transformations can be independent of the pseudoelastic strain when a Lüders-type deformation occurs. This independence is artificial.
3. The transformation-strain dependence of pseudoelastic hysteresis is attributed to the effect of the polycrystallinity of a material. It is suggested that, in a polycrystalline specimen, the formation of a fully oriented martensite in each grain *via* a stress-induced martensitic transformation is accompanied by an internal plastic deformation. The plastic deformation is needed as a coordination mechanism for the deformation mismatch among the preferential variants in neighboring grains. The macroscopic stress hysteresis measured from stress-strain curves is a statistic measure of the resistance to transformation phase–boundary movement and the resistance to the production of this internal plastic deformation.

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