Effect of mechanical cycling on the stress–strain response of a martensitic Nitinol shape memory alloy

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ABSTRACT

An experimental investigation into the ambient temperature, load-controlled tension–tension fatigue behavior of a martensitic Nitinol shape memory alloy (SMA) was conducted. Fatigue life for several stress levels spanning the critical stress for detwinning was determined and compared with that obtained on an alloy similar in composition but in the austenitic state at room temperature. Results show that the fatigue life of the pseudo-plastic alloy is superior to superelastic shape memory alloy. The stress–strain hysteretic response, monitored throughout the fatigue loading, reveals progressive strain accumulation with the cyclic loading. In addition, the area of hysteresis and recoverable and frictional energies were found to decrease with increasing number of fatigue cycles. Post-mortem characterization of the fatigued specimens through calorimetry and fractography was conducted in order to get further insight into the fatigue micromechanisms. These results are discussed in terms of reversible and irreversible microstructural changes that take place during cyclic loading. Aspects associated with self-heating of martensitic alloy undergoing high frequency stress cycling are discussed.

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1. Introduction

The Ni–Ti shape memory alloy (SMA), Nitinol, is widely used in biomedical, structural and aerospace engineering [1–5], owing to its superior performance over other SMAs. SMAs exhibit two characteristic effects, viz., shape memory effect (SME) and superelasticity (SE) or pseudoelasticity (PE). In principle, SME is associated with reversible thermo-elastic martensitic transformation between the austenite (parent) and the martensite (product) phases. This transformation can be induced not only by cooling from the austenitic phase, but also by stressing it. Generally, from a macroscopic perspective, SME is realized either in a three-step transformation involving austenite, twinned martensite and detwinned martensite, or in a two-step process involving austenite and stress-induced (or detwinned) martensite. Fig. 1 schematically illustrates these transformation pathways. Austenite typically transforms into twinned or self-accommodating martensite. On the other hand, the detwinning of martensite upon the application of the stress involves the growth of one variant of the twinned martensite at the expense of an adjacent unfavorably oriented variant (a stress-induced reorientation of martensite) or the formation of stress preferred martensite directly from the austenite [6,7]. It may be noted here that while the transformation between the austenite and twinned martensite and austenite and detwinned martensites are reversible, detwinned martensite cannot be transformed to twinned martensite upon unloading. The latter can happen only through the formation of austenite, i.e., by heating the alloy to above the austenitic finish temperature, $A_f$.

Formation of detwinned martensite results in large strains (typically up to 7%), which is shown to be a result of detwinning and a stress-induced twin boundary migration within a martensite variant. Most of the applications of SMAs exploit this large transformation strain and/or the large stresses associated with these transformations. While the quasi-static tension and compression properties of Nitinol are reasonably well characterized, its fatigue behavior is not. Note that for repeatable and reliable performance of SMA devices, it is essential that the alloy exhibits good mechanical and functional stability under cyclic loading. This assumes greater significance in the case of SMAs wherein the thermo-mechanical response is generally highly path or history dependent. Some irreversible processes are inevitable during the transformation, which in turn accumulate during cyclic loading. These lead to changes in transformation characteristics such as the transformation temperatures and the transformation stress. For instance, the stress–strain response corresponding to the first cycle is markedly different from the response in say, the 100th cycle. These aspects need detailed examination before SMAs can be deployed in reliability-sensitive applications.
The deformation response of SMAs during thermal cycling has been examined in detail and it is well known that Nitinol possesses good fatigue resistance in strain controlled mode. In fact, it has been reported that the fatigue resistance of martensitic Nitinol is higher than conventional metallic materials [8,9]. Though extensive research [8,10–18] has been carried out on fatigue behavior of SMAs, only few studies [19–21] have analyzed the effect of mechanical cycling in the martensitic and R phase transformation range on the overall macroscopic behavior of NiTi SMAs. Keeping this in view, mechanical fatigue experiments on a NiTi alloy that is in the martensitic condition at room temperature were conducted and reported in this paper. Since detwinning that occurs due to the application of stress is non-reversible (unless when taken through austenitic condition), the prime objective of this work is to evaluate the role of irreversible processes that occur during detwinning under fatigue loading on the life of Nitinol. Throughout the fatigue experiments, the stress–strain hysteresis was monitored in an attempt to evaluate the progressive accumulation of damage and other associated effects. These results are compared with similar fatigue experiments conducted on an austenitic Nitinol, which are reported in an earlier paper [22].

2. Materials and experiments

A Nitinol alloy block with a composition of 55.90 wt.% Ni was obtained from Special Metals Corporation, NY, USA, in hot forged and annealed condition. “Dog bone” type tensile specimens with a 60 mm gage length, 5 mm width and 2 mm thickness were electro-discharge machined (EDM) from the block and were annealed at 600 °C (which is slightly above the recrystallization temperature of the alloys) for 30 min in vacuum-sealed quartz tubes, to completely recrystallize the alloy [23,24]. The EDM cut specimens were not electro-polished to see the sole effect of EDM cutting on the surface, as it is the most efficient method for creating structures of high aspect ratio that has low surface roughness.

Differential scanning calorimetry (DSC) results, plotted in Fig. 2, show that the transformation temperatures for the annealed alloy in free stress state are as following: austenitic start, \( A_s = 68 ^\circ C \), austenitic finish, \( A_f = 105 ^\circ C \), martensitic start, \( M_s = 63 ^\circ C \), and martensitic finish, \( M_f = 29 ^\circ C \). This confirmed that the material is in the martensitic state at the fatigue testing temperature (room temperature, \( \sim 22 ^\circ C \)) and is fully recrystallized exhibiting almost identical transformation behavior of single stage B2-B19’ reported by Mahmud et al. [25].

An MTS servo-hydraulic universal testing machine equipped with data acquisition system and hydraulic pressure grips was used for the fatigue tests. Tension–tension fatigue cycling was conducted with a stress ratio, \( R \) (ratio of the minimum to maximum stresses of the fatigue cycle) of 0.1. The frequency was maintained at 5 Hz. Each specimen was cycled in load-controlled mode at fixed stress amplitude, \( \Delta \sigma = (\sigma_{\text{max}} - \sigma_{\text{min}}) \), where \( \sigma_{\text{max}} \) and \( \sigma_{\text{min}} \) are the maximum and the minimum stresses of the sine-wave fatigue cycle and the stress–strain loops were monitored till fracture. An extensometer with a gage length of 25 mm was used for measuring the strain and care was exercised so as to make sure that it did not slip during fatigue loading. All the fatigue tests were conducted at room temperature (RT \( \sim 22 ^\circ C \)).

To aid the understanding of the change in transformation temperatures and micromechanism during mechanical cycling, DSC was employed for characterizing the as-annealed and fatigued specimens. The transformation behavior was analyzed on \( \sim 15 \) mg specimens (carefully machined from the gage sections of the fatigue-fractured specimens using a slow speed diamond saw) using a Mettler–Toledo differential scanning calorimeter. The samples were first heated from room temperature to 150 °C and then cooled to \( -50 ^\circ C \) before heating again to RT at a heating/cooling rate of \( 5 ^\circ C/\text{min} \).

3. Results and discussion

3.1. Mechanical testing

Quasi-static room temperature tensile stress–strain response of the martensitic alloy is shown in Fig. 3. It shows a distinct plateau with the detwinning initiation stress, \( \sigma_{\text{DT}} \), of \( \sim 180 \) MPa below which it is essentially elastic. Above \( \sigma_{\text{DT}} \), the detwinning process of martensite variants occurs [26], which generally involves considerable dissipation of energy through the relative movements between variant interfaces during their reorientation and growth. In polycrystalline SMAs, detwinning can be broadly seen to occur in two stages. Initially, in each twinned martensitic plate, the marten-
Fig. 3. Quasi-static, room temperature tensile stress–strain response of the martensitic Nitinol alloy used in the present study.

sitic variant less favorably oriented to the applied stress detwins leading to large transformation strains. This process is usually seen to occur at a constant stress level and hence results in a plateau in the stress–strain response. However, it should be noted that not all detwinned variants have the same orientation with respect to the loading axis. Hence, due to increase in load, the detwinned variants reorient more favorably to the applied stress, causing additional transformation strain. This is called reorientation. (It may be mentioned that in some literature the term ‘detwinning’ or ‘reorientation’ is used to denote both the stages.) After \( \sim 270 \) MPa, an elastic response is seen again till the onset of conventional yielding due to the plastic deformation of detwinned martensite.

The tensile stress–strain responses of the Nitinol alloy during fatigue loading with the prescribed \( \sigma_{\text{max}} \) of 300 and 500 MPa are shown in Fig. 4a and b, respectively. In the load-controlled fatigue tests, about 10–20 cycles are necessary for the closed-loop servo control to stabilize, with most of stabilization occurring by \( \sim 10 \)th cycle. This is the reason for the stress range in these first few cycles being lower than the prescribed values; instead the maximum stress of the cycle continues to rise with each further cycle. Another interesting aspect is the hardening observed, which is largely due to the relatively high rate of loading (approximate strain rates of \( \sim 10/\text{s} \)) experienced during fatigue [27].

The formation of hysteresis loops, even after the stabilization of the cyclic loads, may be explained on the basis of dissipation of the non-chemical irreversible energy that arises due to the interfacial friction between adjacent martensite variants during their reorientation process [26,28]. During thermal cycling, hysteresis occurs as a result of the frictional resistance experienced by the transformation phase boundaries. The growth of certain martensite variants due to applied stress and their reorientation lead to gradual dislocation accumulation at the interface between different martensite colonies [29,30]. The hardening observed during cyclic loading is possibly related to this effect [30]. Macroscopically, the effect of these changes in the microstructure manifest as accumulated (remnant) strain at the end of each cycle. Fig. 5 shows the stabilized stress–strain hysteresis loops generated at various stress amplitudes, all at the fatigue cycle number, \( N = 100 \). At all the stress ranges, the loops have relatively narrow width with near-linear loading and unloading response. This is in contrast to austenitic Nitinol, which exhibits elastic deformation followed by non-linear deformation during loading and unloading for the loops generated above \( \sigma = 286 \) MPa [22]. With increasing \( \Delta \sigma \), the hysteresis loops move to the right on the strain axis, indicating higher amounts of strains already accumulated in the SMA.

Fig. 6 shows the effect of different stress amplitude on the accumulated plastic strain, \( \varepsilon_p \), as a function of the number of fatigue cycles. Here, \( \varepsilon_p \) is operationally defined as the deviation of the starting strain of the hysteresis loop after \( N \)th cycle. It is seen that \( \varepsilon_p \) increases rapidly between 10th and 20th cycles before reaching a plateau value that is maintained until failure. The initial several percent of strain result from a combination of stress-induced growth of one martensite variant at the expense of an adjacent, unfavorably oriented one as well as stress-induced reorientation of martensite and twin boundary migration within a martensite variant. These processes lead to the generation of high density of dislocations within the martensite variants and inside twins due to cyclic deformation. It is believed that internal plastic deformation is necessary for martensite reorientation in polycrystalline matrices, due to orientation mismatch among the preferential variants of martensite in neighbouring grains [30]. Therefore, the interface between any two martensite variants becomes highly strained, in turn, making the dislocation mediated deformation dif-
Fig. 5. Steady-state stress–strain hysteresis loops obtained at cycle number, $N$, of 100, each generated with various stress amplitudes.

Fig. 6. Variations in the accumulated plastic strain, $\varepsilon_p$, vs. number of fatigue cycles, $N$, at different levels of maximum stress, $\sigma_{\text{max}}$.

Fig. 7. Hysteresis width at the mean stress level, $\varepsilon_H$, vs. number of fatigue cycles, $N$, at different maximum stress level, $\sigma_{\text{max}}$.

Fig. 8a and b that both $E_1$ and $E_2$ decrease with increasing $N$, which can be explained as follows. Initially, a number of martensitic variants are present within the SMA. Out of these, those which are favorably oriented grow at the cost of other by twin boundary movement, leading to a general reduction in the number of moving boundaries with stress cycling. It can be observed that the rate of decrease of $E_2$ with $N$ at a given stress amplitude is not as fast as that of $E_1$. The decrease in $E_1$ suggests that the martensite damping capacity decreases with increasing number of cycles. Compared to austenitic alloy, martensitic alloy has higher recoverable energy for a given load level. This could be due to the reduced hysteresis (reduced $E_1$) and for the same strain level, the recoverable contribution will be higher. However, the similar to other features discussed above, $E_2$ stabilizes beyond ~20 cycles. The energy dissipated in 100th cycle, $E_{100}$, is plotted as a function of $\sigma_{\text{max}}$ in Fig. 9, in order to examine the effect of stress amplitude on energy dissipation during fatigue. It shows that $E_{100}$ increases exponentially with $\sigma_{\text{max}}$. This is because with increase in stress range the volume fraction of detwinned martensite increases and the proportional energy loss of the detwinned martensite is only half of the twinned martensite as observed by Huang and Lim [31].

In Fig. 10, a comparison of fatigue strain accumulation in martensitic and austenitic Nitinol alloys (reported in an earlier paper by us [22]), both subjected to stress-controlled fatigue loading, was made. For this purpose, the accumulated plastic strain at 100th cycle, $\varepsilon_{p,100}$, was chosen as the. It is seen that both the alloys show essentially similar strain accumulation benchmark characteristics. Note that strain accumulation in the initial few cycles is due to shakedown in untrained SMAs. Fig. 11a and b shows the hysteresis width and the recoverable deformation energy at 100th cycle ($\varepsilon_{H1,100}$ and $E_{1,100}$) plotted against maximum stress of the fatigue cycle, for both austenitic and martensitic alloys. Both these figures show that, for all stress amplitudes, both $\varepsilon_{H1,100}$ and $E_{1,100}$ in the austenitic alloy are significantly higher than the corresponding values recorded for the martensitic alloy. Some amount of reversible SIM transformation (the extent of which depends on stress level) in austenitic alloy leads to larger hysteresis width (larger strains) and also larger recoverable deformation energy compared to lesser strains and irreversible nature of detwinning transformation that occurs in martensitic alloy.

The number of cycles to failure, $N$, is plotted against the stress amplitude, $\sigma_3$ in Fig. 12. It is seen that for $\sigma_3 > 300$ MPa, $N$ of martensitic Nitinol is an order of magnitude higher than that of the austenitic alloy. This is despite the fact that stress required for the onset of non-linear deformation (detwinning stress in the case of martensite and stress required for inducing austenite-to-martensite transformation in the case austenite) is significantly smaller for the martensitic alloy (~180 MPa) as compared to that.
of the austenitic alloy (~260 MPa [21]). A possible factor for the lower fatigue resistance of the austenitic alloy is the stabilization of stress-induced martensite (SIM) upon fatigue loading, which makes the alloy biphasic. Another factor could be that martensite microstructure is more tolerant and softer for fatigue induced damage propagation and failure. It is widely recognized that higher crack growth rates and lower fatigue threshold value are observed in two phase microstructures [32–34]. In the context of austenitic Nitinol, Wagner et al. [35] have shown that it is possible to erase, at least partially, the strain that accumulates during fatigue by recourse to a simple annealing treatment above \( A_f \). It would be interesting to examine if such a possibility also exists for the martensitic Nitinol.

3.2. Structural characterization

DSC was performed on annealed and fatigue-fractured samples in order to examine any changes in the transformation behavior due to fatigue. The scans obtained during cooling and heating cycles are shown in Fig. 13a and b, respectively. All the fatigued samples show essentially an identical response. The exothermic peak seem to be unaffected by fatigue; only slight broadening of the peak with \( M_s \) increasing and the martensitic finish temperature \( M_f \) decreasing (Fig. 13b). However, a significant difference is noted in the transformation behavior upon fatiguing. The heat flow trace in DSC scan during heating shows that \( A_s \) temperature increased for all fatigued samples as compared to the annealed sample. The critical temperature for the reverse transformation of the deformed martensite of the first heating are consistent with the results of the previous studies [29,36]. This is because the martensitic reorientation in polycrystalline NiTi alloys causes thermal stabilization of the martensite, resulting in an increase in the critical temperature for the reverse transformation. For samples fatigued at low stress level (at \( \sigma_{\text{max}} = 350 \) MPa) single peak (single step transformation) was observed which started broadening with increase in stress level at (\( \sigma_{\text{max}} = 400 \) MPa and \( \sigma_{\text{max}} = 500 \) MPa) and finally splits into two (two-step transformation) at \( \sigma_{\text{max}} = 550 \) MPa stress level as shown in Fig. 13a. The splitting of the peak into two during heating indicates two-step transformation. One possible reason for this could be the differences in the residual stresses in the material. Higher residual stresses necessitate higher temperatures to induce reverse transformation.

![Fig. 8](image8.png)

Fig. 8. (a) Area of the stress–strain hysteresis loop, \( E_1 \), and (b) recovered deformation energy, \( E_2 \), plotted as a function of the number of fatigue cycles, \( N \), for different maximum stress, \( \sigma_{\text{max}} \).

![Fig. 9](image9.png)

Fig. 9. Area of the stress–strain hysteresis loop at \( N = 100 \), \( E_{1,100} \), plotted as a function of the maximum stress applied during the loading cycle, \( \sigma_{\text{max}} \).

![Fig. 10](image10.png)

Fig. 10. Maximum accumulated strain, \( \epsilon_{p,100} \), plotted as a function of a the maximum stress applied during the loading cycle, \( \sigma_{\text{max}} \). Note that \( \epsilon_p \) plotted here is taken only after \( N = 100 \).
Fig. 11. (a) Hysteresis width, $\varepsilon_{H,100}$, and (b) recoverable deformation energy, $E_{1,100}$, which are measured at 100th cycle, plotted as a function of the maximum stress applied during the loading cycle, $\sigma_{\text{max}}$.

Another possible explanation could be the following. Cycling at high frequency (5 Hz) and large $\Delta \sigma$ can lead to self-heating of the material due to repeated cyclic deformation involving dissipative mechanisms like interface friction [34,37–40]. The factors responsible for self-heating are the frequency of loading, amount of phase transformation and associated hysteresis related to $\Delta \sigma$ and the ambient conditions. Large to severe self-heating (exceeding 40 K) is reported in the cases of fatigue at frequencies greater than 1 Hz and at stress levels causing large transformation strains [39]. It is reported that fatigue in martensitic condition is likely to induce more self-heating compared to austenitic alloy. One of the reasons for this is the lack of endothermic reverse transformation in martensite during fatigue vis-à-vis austenite. It may be noted that in austenitic fatigue, the forward (A $\rightarrow$ M) exothermic transformation tends to increase the sample temperature, while the reverse (M $\rightarrow$ A) endothermic transformation offsets the temperature increase. This self-heating (quite high due to cycling at 5 Hz) introduces several complexities in the fatigue behavior of martensitic alloys. Some of these aspects are discussed below and they need further detailed investigations to obtain better understanding.

Fig. 12. Stress amplitude, $\Delta \sigma$, vs. number of cycles to failure, $N_f$, for martensitic Nitinol alloy examined in this study. For comparison purposes, similar data, but obtained from Ref. [22] on an austenitic Nitinol alloy is also plotted.

Fig. 13. DSC scans obtained during (a) cooling and (b) heating of the samples that were extracted from the gage potions of fatigue tested specimens.
Fig. 14. Schematic diagram showing temperature rise during fatigue loading. Depending on the stress range and the nature of dissipative processes, $T_{\text{max}}$ can vary significantly, possibly exceeding $A_f$ from being less than $A_s$ to above $A_f$. Further, it may be noted that there will be temperature variations in the specimen. This introduces complexities in the fatigue response. To illustrate several scenarios that occur due to self-heating and thermal gradients, a schematic stress-temperature phase diagram showing phase transformation (active) and inactive regions is shown in Fig. 15. It may be noted that this is only schematic and represents an idealized situation. Nonetheless, from Fig. 15, it is clear that depending on $T_{\text{max}}$, several possible transformation scenarios exist for the same stress amplitude. If the temperature exceeds $A_s$ (for higher $\Delta \sigma$ or for large $N$), then during the unloading segment of stress cycle, the stress-induced martensite transforms into austenite. For this austenite, some amount of reversible SIM (superelastic) transformation can occur leading to formation of phase inhomogeneities. If the temperature exceeds $A_f$, then the austenite that is formed does not undergo superelastic loading cycles due to the stresses being less than the stress necessary for SIM formation. Thus, due to temperature gradients, there will be significant material inhomogeneity with regions of detwinned martensites ($T < A_s$) and austenite and different variants of stressed martensites. Due to applied mechan-

Fig. 15. Schematic diagram showing phase transformation regions in the stress-temperature space. $M_t$ is the twinned martensite, $M_d$ is the detwinned or stress preferred martensite and $A$ is the austenite. The applied stress range is also shown. $A_s$ and $A_f$ are the austenite start and finish temperatures, respectively.

Fig. 16. Differences in fracture features at different locations for two stress levels. (a) and (b) are from the sample fatigued at $\sigma_{\text{max}} = 300$ MPa whereas (c) and (d) are from that with $\sigma_{\text{max}} = 500$ MPa. While (a) and (c) are obtained from the edges of the fractured specimens where cracks have initiated, (b) and (d) are from the centers of the specimens which is a fast fracture regime.
tical cycling, incoherency in the form of severe strain gradients can occur at phase boundaries. This will be a major contributing factor for the fatigue degradation.

Formation of cracks leads to steep stress gradients which in turn causes superelastic cycling near the crack tips at temperatures higher than $A_s$. This could be accompanied by plastic (irreversible) deformation of the stress-induced martensite, resulting in worsening of strain gradients, thereby causing formation and growth of interfaces and secondary cracking. Formation of large number of secondary cracks is evident from the fractographs shown in Fig. 16. This reduces the fatigue life of the material. Upon fracture, relaxation of stress on the fractured surface could lead to formation of twinned martensite when the austenite cools to room temperature. However, it may be noted that there will also be martensite under residual stress after failure of the specimen. This is evidenced by the twin peaks in the DSC cooling curves for higher stresses. Peak corresponding to higher transformation temperatures indicate martensite under higher internal stress.

For samples stressed at lower stress amplitude, large number of cycles at 5 Hz causes significant self-heating which leads to formation of significant amount of austenite. This will remain austenite since the load is not enough for formation of SIM at higher temperatures. Formation of multiple austenite–martensite interfaces and cyclic loading leads to generation and growth of fatigue cracks, in turn, leading to failure. Hence the endurance limit of the martensitic alloy is much lower than that of the austenitic alloy, partly attributable to the lower detwinning stress in martensitic alloy. Comparison of DSC cooling trace of the 350 MPa maximum stress and the DSC of the annealed sample shows that the martensite that is formed is after cooling to room temperatures is relatively stress free leading to a single peak corresponding to twinned martensite.

Differences in features on the fracture surface can be seen from Fig. 16a–d. During self-heating, regions near the edge are at lower temperatures compared to those near the center due to heat loss to the surroundings. This thermal gradient seems to influence the fracture features. For instance, the center portions being more biphasic, show more dimple features suggesting static overload due to larger strain incoherence. Lower temperatures near the edges seem to facilitate more fatigue failure. However, more detailed studies on the microstructure of the fatigued samples are necessary to confirm different variants of martensite. Additionally, cycling of martensitic alloy at lower rates is recommended to minimize the impact of self-heating. It appears that stressing at less than 0.5 Hz and use of appropriate convective cooling to take away the heat ensures that the martensite does not reverse transform into austenite during cyclic loading.

4. Summary

Stress-controlled tension–tension fatigue was performed on martensitic NiTi alloy to investigate its fatigue behavior. Stress–strain hysteresis was monitored throughout the test to assess the effects of several irreversible mechanisms that occur during irreversible detwinning and reorientation in martensitic alloys subjected to cyclic mechanical loading. Post-mortem studies including DSC and fractography were conducted to understand the degradation due to fatigue. These investigations show that in general, Nitinol in the martensitic state at room temperature has an order of magnitude higher fatigue life than the austenitic state. The frictional energy lost due to cyclic movement of interface during the martensitic reorientation is a major contributing factor for the hysteresis profile. After stabilization of stress during mechanical cycling, hysteresis width, recoverable energy and frictional energy for each number of cycles decrease markedly for early number of cycles and then approach a saturated value. Significant self-heating associated with high frequency and large number of cycles introduces several complexities in analyzing the martensitic fatigue. Significant self-heating associated with relatively high frequency and large number of cycles introduces several complexities in analyzing the martensitic fatigue. Cycling at lower frequencies (<1 Hz) with appropriate cooling may be necessary to adequately characterize fatigue in martensitic SMA.

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References