Microstructural and Mechanical Stability of a Ti-50.8 at.% Ni Shape Memory Alloy Achieved by Thermal Cycling with a Large Number of Cycles

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Abstract: The influence of thermal cycling (TC) with a large number of cycles on the microstructure, the parameters of martensitic transformations (MTs), and the mechanical properties of a Ti-50.8 at.% Ni shape-memory alloy in coarse-grained (CG) and ultrafine-grained (UFG) states was investigated. The effect of microstructural and mechanical stability was found in both coarse-grained and ultrafine-grained states starting from the 100th cycle of martensitic transformations. In addition, an unusual temperature change was observed in martensitic transformations occurring with the formation of an intermediate R phase.

Keywords: TiNi alloy; thermal cycling; ultrafine-grained structure; microstructural and mechanical stability

1. Introduction

The phenomenon of shape memory is characteristic of some alloys based on titanium, iron, copper, or manganese. Particular attention to materials with the shape memory effect (SME) is associated with their ability of recovering significant inelastic deformations when heated. The most famous representative of such materials is titanium nickelide. Its unique properties are widely used in world industries and in medicine [1,2]. The SMEs in titanium nickelide are caused by the B2–B19’ thermoelastic martensitic transformations (MTs) occurring in the temperature range close to room temperature [1–4]. The cycle of martensitic transformations upon cooling and heating leads to the generation of dislocations in the crystal lattice. Understanding the nature of the influence of multiple cycles of “cooling and heating” below and above the points of martensitic transformation—thermal cycling (TC)—on the structure and properties of materials is of great importance, particularly for TiNi alloys and their products. The phenomenon of phase hardening (PH)—the accumulation of dislocations during martensitic transformations—does not seem trivial in the case of martensitic transformation with a reversible motion of martensitic boundaries. The term “thermoelastic transformation” in the strict sense does not imply irreversible changes in the structure. At the same time, in real metallic materials, including TiNi alloys, a certain increase in the dislocation density occurs during multiple MT cycles, which, in turn, is accompanied by a change in the martensitic transformation temperature and an increase in the dislocation yield strength of the alloys under mechanical loading [5–7].

The design of products with SME makes certain demands on the physico-mechanical and functional properties and their stability. The properties in alloys with shape memory can be further
improved by forming an ultrafine-grained (UFG) state using severe plastic deformation (SPD) methods, in particular, equal channel angular pressing (ECAP) [8–13]. Because TiNi system alloys are the most common in technological applications and have the best set of properties among alloys with a shape memory effect, the effect of thermal cycles (TCs) on their structure and properties has been studied for many years. In TiNi alloys, the transformation of B2 into B19′ is characterized by the incompatibility of lattice deformation, which contributes to the emergence of local stresses at the phase boundary, and stress relaxation leads to the accumulation of plastic deformation and, as a consequence, irreversible changes in the kinetics of martensitic transformations with each thermal cycling cycle [14]. The first works [5,6] were devoted to the influence of TCs on the structure and characteristic temperatures of martensitic transformations, and the mechanical characteristics in the TiNi alloy. It was previously shown that thermal cycling through the interval of martensitic transformations leads to a change in the staging of the transformation [15–19]. Alloy Ti₅₀Ni₅₀ undergoes B19′ → B2 transformation upon cooling. However, after several thermal cycles during cooling, the alloy begins to experience a multi-stage B2→R→B19′ transformation. At the same time, other studies report slightly different dependences of the transformation temperatures upon thermal cycling under an applied load [20,21]. There are additional studies of the “thermocyclic training” of TiNi alloys to enhance memory effects [20,22–24]. However, the studies were carried out mainly on alloys in a coarse-grained (CG) state, or in states with a small degree of deformation, and there is a limited number of studies on the processes of accumulation of dislocations and change in properties during thermal cycling of TiNi alloys in UFG and nanocrystalline (NC) states [25]. In addition, the conducted studies did not determine how many cycles were required to obtain optimal characteristics of the properties in these alloys. The studies in this work were aimed at determining the optimal number of thermal cycles necessary to obtain stability of the structure and the physico-mechanical properties of the TiNi alloy in coarse-grained and ultrafine-grained states.

2. Materials and Methods

As a research material, a two-component alloy of the TiNi system was chosen—the stoichiometric alloy Ti-50.8 at.% Ni enriched with nickel, manufactured by MATEK-SMA Ltd. (Moscow, Russia). It has a bcc lattice ordered by type B2 (CsCl) and a phase enriched with nickel Ti₂Ni₃ [1,4], and the chemical composition of the alloy is presented in Table 1. To obtain a solid solution and eliminate processing history, quenching was carried out from the homogeneity region (heating at a temperature of 800 °C in a Nabertherm furnace for 1 h) into the water. The average grain size of the hardened alloy was about 20 ± 2 μm.

Table 1. Chemical composition of the alloy of titanium nickelide, % (by atomic).

<table>
<thead>
<tr>
<th>Main Elements</th>
<th>Impurity</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ni</td>
<td>Remainder</td>
</tr>
</tbody>
</table>

To carry out the deformation by the ECAP method, the equipment of the Ufa State Aviation Technical University (USATU) design in isothermal mode was used. To form the UFG structure, quenching samples of cylindrical TiNi alloys (Ø20 mm, length 100 mm) were subjected to 8 passes along the Bc route at 450 °C, and the channel intersection angle (φ) was 120° [9]. Thermal cycling of the samples in different initial states was carried out as follows: the samples were successively immersed in liquid nitrogen (−196 °C), then they were heated to a temperature of 150 °C, which is actually lower and higher than the temperatures Mₐ of direct and Aₙ reverse martensitic transformation. The number of heating–cooling thermal cycles ranged from 0 to 250. The thickness of the samples subjected to TC in the section was less than 1 mm to ensure their rapid heating and cooling. The exposure time at the heating and cooling temperatures was 5 min [25]. Quantitative and qualitative analyses of the initial structure were carried out using an OLYMPUS GX51 metallographic microscope. To detect the microstructure, an etchant with a composition of 60% H₂O + 35% HNO₃ + 5% HF was used. Using the random secant method, the sizes of structural elements were calculated.
X-ray diffraction studies of the samples were carried out on a Rigaku Ultima IV diffractometer (\(U = 40\) kV and \(I = 35\) mA) at room temperature in the angle range \(2\theta = 30^\circ–120^\circ\). The main structure parameters were determined by the Rietveld method using the Materials Analysis Using Diffraction (MAUD) program. The dislocation density was calculated by processing X-ray diffraction data using MATLAB software. The formula was used to calculate the density of dislocations [8]:

\[
\rho = 2\sqrt{3} \frac{<\epsilon^2>^{1/2}}{D b},
\]

where \(<\epsilon^2>^{1/2}\) represents the microdistortions, \(D\) is the average grain size, and \(b\) is the Burgers vector.

The fine structure of the material was studied at room temperature using a JEOL JEM-2100 transmission microscope (“JEOL Ltd.”, Tokyo, Japan) with an accelerating voltage of 200 kV. Samples for thin foils cut by the electro-erosion method were made by double-sided jet electrolytic polishing using a Tenupol-5 device (“Struers”, Copenhagen, Denmark) in a solution of 10% perchloric acid (HClO\(_4\)) and 90% butanol (CH\(_3\)(CH\(_2\))\(_3\)OH).

The average size of structural elements (grains, subgrains, martensitic twins) was estimated using the “GrainSize” software package by measuring chord lengths, the relative measurement error of which did not exceed 5%. In this work, the calorimetric testing of the material was carried out on a Mettler Toledo high-sensitivity differential scanning calorimeter (“Mettler Toledo”, Columbus, OH, USA) on samples weighing up to 50 mg (diameter 3.5 mm, thickness 0.5–0.7 mm), and the change in heat flux was studied during cooling and heating in the temperature range from \(-196^\circ\) C to 150 \(^\circ\)C at a rate of 10 \(^\circ\)C/min. The temperatures of the beginning (\(M_r\) and \(A_r\)) and the end (\(M_f\) and \(A_f\)) of the direct and reverse transformations were determined by standard tangent methods (ASTM 2004-05).

The microhardness \(H_v\) in this work was determined by the Vickers method on a Micromet 5101 instrument with a diamond indenter (“Buehler”, Lake Bluff, OH, USA). Mechanical tensile tests of small flat specimens in compliance with all dimensional ratios with a working part of 1 \(\times\) 0.25 \(\times\) 4 mm were carried out with a strain rate of 1 \(\times\) 10\(^{-3}\) s\(^{-1}\) on a special installation of the USATU design at room temperature. According to the test results, the strength characteristics (phase yield stress \(\sigma_{m}\), dislocation yield strength \(\sigma_{YS}\), ultimate tensile strength \(\sigma_{UTS}\)) and ductility (elongation, \(\delta\)) were determined. The difference between the dislocation and phase yield limits allows one to estimate the stress \(\sigma_{reac} = \sigma_{YS} - \sigma_{m}\) (estimated reactive stress) [4,25]. The length of the plateau at the phase yield stress stage was adopted as an estimate of the recovery strain \(\epsilon_{rec}\) (Figure 1). The values of the conditional stresses of phase yield \(\sigma_{m}\), dislocation yield stress \(\sigma_{YS}\), and ultimate tensile strength \(\sigma_{UTS}\) are calculated as average statistical values for three samples.

![Figure 1. Diagram of a tensile stress indicating mechanical characteristics.](image)

3. Results

The phase composition of TiNi alloys significantly affects the microstructure, functional and mechanical properties, as well as the processes that can occur during thermal cycling. It is known that with an increase in the Ni content, the tendency to phase hardening and, accordingly, the
sensitivity of the structure to the accumulation of defects decrease. In this paper, the features of the influence of TCs on the characteristics of the alloy are considered on the example of an understated Ti-50.8 at.% Ni alloy up to 250 cycles.

3.1. Study of the Microstructure of the Ti-50.8 at.% Ni Alloy under Various Conditions

3.1.1. Structure in the Coarse-Grained State

In the initial state, after quenching, the alloy at room temperature has a predominantly equiaxial structure of the B2 phase (austenite) with an average grain size of about 20 ± 5 μm (Figure 2a). Moreover, globular inclusions 0.5–1 μm in size are observed in the structure inside and at the grain boundaries. Optical microscopy (OM) and scanning electron microscopy (SEM) (Figure 3) fail to accurately assess changes in the structure after repeated thermal cycling (Figure 2b); therefore, the TEM method was used (Figure 4).

![Figure 2](image1.png)
![Figure 3](image2.png)

**Figure 2.** Microstructure of the Ti-50.8 at.% Ni alloy obtained by optical microscopy (OM): (a) in the initial coarse-grained (CG) state, (b) after thermal cycling with a maximum number of cycles \( n = 250 \).

**Figure 3.** SEM image of the microstructure of the Ti-50.8 at.% Ni alloy in the coarse-grained state: initial state (a), 250 thermal cycles (b).

According to the obtained TEM data in the CG state without thermal cycling in the microstructure of the alloy, grain boundaries and triple-grain junctions free of dislocations are observed (Figure 4a). After thermal cycling in the temperature range MT B2→B19′ with a number of cycles equal to 50, developed dislocation clusters are present that form the so-called “dislocation
forest”. The average grain size decreased insignificantly and became about 18 ± 3 microns in size. With an increase in the number of thermal cycles in the structure, the formation of large clusters of dislocations and dislocation walls, which are formed during phase hardening, was observed. This was first noticeable near the grain boundaries (Figure 4b). In this case, there is a broadening of the extinction contours, which is also associated with an increase in the level of internal stresses and distortions of the crystal lattice. The average grain size at the maximum number of cycles compared with the initial value decreased by about 45% (the assessment was carried out according to OM and SEM). Thermal cycling with the maximum number of thermal cycles preserves the dislocation structure in the form of clusters and irregular walls and dislocation tangles (Figure 4f).

Figure 4. TEM images of the microstructure of the Ti-50.8 at.% Ni alloy in a coarse-grained state with different numbers of thermal cycles: (a) $n = 0$, (b) $n = 50$, (c) $n = 100$, (d) $n = 150$, (e) $n = 200$, (f) $n = 250$. 
Figure 5 presents structures after \( n = 200 \) and \( n = 250 \) cycles with fields with extinction contours, the width of which reaches \( 150 \pm 20 \) nm, which may indicate a high density of defects accumulated during multiple martensitic transformations.

![Figure 5](image)

**Figure 5.** Microstructures of the Ti-50.8 at.% Ni alloy in a coarse-grained state after thermal cycling with \( n = 200 \) (a) and \( n = 250 \) (b) with fields of the extinction contours.

Figure 6 shows a graph of the average grain size versus the number of cycles.

![Figure 6](image)

**Figure 6.** Graph of changes in the average grain size with an increase in thermal cycles in the CG state.

In addition, starting from 100 cycles of martensitic transformation, sections of the structure are observed in which packets of martensitic plates are visible (plate thickness in the range of 50–300 nm). Present in individual plates at high magnifications were composite (001) B19’ nanotwins of type I with a width of up to 5 nm. The formation of nanotwins is probably associated with saturation of the structure after a certain number of cycles.

3.1.2. Microstructure in the Ultrafine-Grained State

After applying the ECAP method, the transformation of the initial coarse-grained structure into an inhomogeneous grain–subgrain ultrafine-grained structure with an increased dislocation density was observed (Figure 7). The microdiffraction pattern corresponds to an ultrafine-grained structure with the presence of diffuse cords.
As the number of martensitic transformation cycles increases, grains are observed in the structure with predominantly nonequilibrium boundaries, which may indicate a highly defective structure (Figure 8). Nevertheless, complex dislocation structures formed in the grains—various accumulations and tangles—and the density of dislocations increased. The average size of structural elements in the UFG state without thermal cycling was $320 \pm 15$ nm, which decreased to $260 \pm 20$ nm after maximum thermal cycling (Figure 9).
3.2. X-Ray Analysis of the TiNi Alloy

To determine the effect of multiple martensitic transformations on the structural characteristics of the Ti-50.8 at.% Ni alloy, an X-ray diffraction analysis was performed at room temperature. X-ray diffraction patterns of the alloy indicate that the main phase in the CG state is B2-austenite (Figure 10). After TC with the maximum number of cycles, the X-ray diffraction pattern indicates a change in the phase composition; instead of peaks on the B2 phase, a doublet peak of phase B19' is observed. A slight broadening of all peaks and a decrease in their intensity were observed (Figure 10). This can
be explained by structural changes in the alloy—an increase in the density of dislocations accumulating during multiple cycles and an increase in internal microdistortions.

The X-ray diffraction pattern of the alloy in the UFG state also corresponds to the B2 phase of austenite (Figure 11). Moreover, there were more lines in this state than in the coarse-grained state. After TC, there was a broadening of the main peaks (Figure 11), which is caused by distortions of the crystal lattice and large values of microdistortions. However, there was no change in the phase composition of the alloy. The presence of only the B2 phase in the X-ray diffraction pattern in the ultrafine-grained state after the maximum number of cycles can be explained by the fact that, in this state, the structure is more homogeneous and the reverse transformation proceeds with the formation of an intermediate R phase with lower transformation energy. This all contributes to the reverse martensitic transformation, and in the UFG state it completely ends. In the case of the coarse-grained state, the structure is more heterogeneous, which affects the heterogeneity of the martensitic transformation; therefore, part of the material may contain a martensitic phase, the presence of which is recorded by X-ray diffraction analysis.

Figure 10. XRD patterns of the Ti-50.8 at.% Ni alloy in the coarse-grained state before thermal cycling and after maximum thermal cycling.

Figure 11. XRD patterns of the Ti-50.8 at.% Ni alloy in the ultrafine-grained state before thermal cycling and after maximum thermal cycling.
Based on the obtained X-ray diffraction data, the following structural parameters were calculated: coherent scattering regions (CSRs), lattice parameter (a), magnitude of the root-mean-square microdistortions of the crystal lattice ($\langle \varepsilon^2 \rangle^{1/2}$), and dislocation density ($\rho$) [26]. The results are shown in Table 2. An analysis of the structural parameters showed that in both states there is a decrease in the CSR values, an increase in internal microdistortions, and an increase in the dislocation density related to them. The dislocation density increased more in the ultrafine-grained state than in the coarse-grained state, which suggests that a higher density of grain boundaries and a smaller grain size contribute to the intensity of defect accumulation.

### Table 2. Structural parameters of the Ti-50.8 at.% Ni alloy.

<table>
<thead>
<tr>
<th>State</th>
<th>Parameters of Structure</th>
<th>Parameter Lattice a, Å</th>
<th>CSR, nm</th>
<th>$\langle \varepsilon^2 \rangle^{1/2} \times 10^{-4}$</th>
<th>$\rho \times 10^{15}$, m$^{-2}$</th>
</tr>
</thead>
<tbody>
<tr>
<td>CG</td>
<td></td>
<td>3.013 ± 0.001</td>
<td>97 ± 2</td>
<td>0.8 ± 0.1</td>
<td>0.5 ± 0.1</td>
</tr>
<tr>
<td>CG + TC</td>
<td></td>
<td>2.895 ± 0.001 (monoclinic)</td>
<td>37 ± 2</td>
<td>2.2 ± 0.1</td>
<td>1.6 ± 0.1</td>
</tr>
<tr>
<td>$\Delta$</td>
<td></td>
<td>0.118</td>
<td>60</td>
<td>1.4</td>
<td>1.1</td>
</tr>
<tr>
<td>UFG</td>
<td></td>
<td>3.011 ± 0.003</td>
<td>35 ± 3</td>
<td>2.7 ± 0.1</td>
<td>5.3 ± 0.15</td>
</tr>
<tr>
<td>UFG + TC</td>
<td></td>
<td>3.013 ± 0.001</td>
<td>19 ± 2</td>
<td>3.4 ± 0.1</td>
<td>7.1 ± 0.1</td>
</tr>
<tr>
<td>$\Delta$</td>
<td></td>
<td>0.002</td>
<td>16</td>
<td>0.7</td>
<td>1.8</td>
</tr>
</tbody>
</table>

$\Delta$ = Parameter difference between the initial state of the alloy and the state after TC.

3.3. Differential Scanning Calorimetry (DSC)

According to the data obtained in the coarse-grained state, during direct martensitic transformation (DMT), one distinct exothermal peak appears on the DSC curves. During the reverse martensitic transformation (RMT), an endothermic peak is observed associated with the appearance of the high-temperature austenitic phase B2 from the martensitic phase B19'. After thermal cycling with $n = 100$ cycles, a peak was observed from the intermediate R phase during direct martensitic transformation and a decrease in the temperatures of martensitic transformations (M_s, A_f). After the maximum number of cycles, a multidirectional change in temperatures was observed, including a slight decrease in the temperatures of the beginning of the direct conversion (M_s) and the end of the reverse (A_i), and an increase in the temperatures of the end of the direct conversion (M_f) and the beginning of the reverse (A_r) (Table 3). In the ultrafine-grained state, a decrease in martensitic temperatures was observed with an increase in the number of cycles from 0 to 250; at the same time, in this state, a transformation with the formation of an intermediate R phase proceeded without thermal cycling, and thermal cycling reduces temperatures—R_s (the temperature of the start of R transformation with direct MT) and R_f (the temperature of the end of the R transformation with direct MT). In this case, the decrease in the temperatures of direct martensitic transformation in the ultrafine-grained state was lower than that in the coarse-grained state. The graphs of the DSC curves of Ti-50.8 at.% Ni alloy are plotted according to Table 3 (Figure 12).

### Table 3. Results of differential scanning calorimetry (DSC) of the Ti-50.8 at.% Ni alloy.

<table>
<thead>
<tr>
<th>States</th>
<th>Number of Cycles</th>
<th>$M_s$, °C</th>
<th>$M_f$, °C</th>
<th>$R_s$, °C</th>
<th>$R_f$, °C</th>
<th>$A_i$, °C</th>
<th>$A_r$, °C</th>
<th>$A^R$, °C</th>
</tr>
</thead>
<tbody>
<tr>
<td>CG</td>
<td>$n = 0$</td>
<td>3.07</td>
<td>-60.11</td>
<td>-</td>
<td>-</td>
<td>-25.07</td>
<td>26.83</td>
<td>-</td>
</tr>
<tr>
<td></td>
<td>$n = 100$</td>
<td>-38.60</td>
<td>-98.32</td>
<td>8.46</td>
<td>-</td>
<td>-29.55</td>
<td>8.47</td>
<td>-</td>
</tr>
<tr>
<td></td>
<td>$n = 250$</td>
<td>-49.63</td>
<td>-70.23</td>
<td>11.62</td>
<td>-29.5</td>
<td>-18.33</td>
<td>-8.14</td>
<td>-</td>
</tr>
<tr>
<td>UFG</td>
<td>$n = 0$</td>
<td>-13.52</td>
<td>-80.96</td>
<td>39.44</td>
<td>2.18</td>
<td>6.11</td>
<td>33.16</td>
<td>-</td>
</tr>
<tr>
<td></td>
<td>$n = 100$</td>
<td>-16.98</td>
<td>-74.27</td>
<td>38.60</td>
<td>2.24</td>
<td>7.83</td>
<td>36.65</td>
<td>-</td>
</tr>
<tr>
<td></td>
<td>$n = 250$</td>
<td>-59.72</td>
<td>-83.09</td>
<td>15.10</td>
<td>-32.07</td>
<td>-19.87</td>
<td>-9.87</td>
<td>7.97</td>
</tr>
</tbody>
</table>

$A^R$ = Temperature of the end of the reverse transformation with an intermediate R phase.
3.4. Mechanical and Functional Characteristics of Ti-50.8 at.% Ni Alloy during Thermal Cycling

In both CG and UFG states of the Ti-50.8 at.% Ni alloy, the values of microhardness as a result of multiple martensitic transformations increased slightly compared to the state before thermal cycling (Figure 13). However, in the first 50–100 cycles, a more intense increase in microhardness is characteristic, then the values stabilize and saturation occurs after 100 cycles of transformations.

![DSC curves of the Ti-50.8 at.% Ni alloy in CG (a) and UFG (b) states.](image)

**Figure 12.** DSC curves of the Ti-50.8 at.% Ni alloy in CG (a) and UFG (b) states.

The results of mechanical tensile tests are presented in Figures 14 and 15 for CG and UFG states. The functional characteristics determined from the analysis of mechanical tensile tests are presented in the form of graphs of the dependence of the estimated reactive stress ($\sigma_{\text{reac}}$) and plateau length at the phase yield stage, used as an estimate of the recovery deformation ($\varepsilon_{\text{rec}}$) on the number of thermal cycles (Figure 16).

![Graph showing the dependence of microhardness on the number of thermal cycles in various states.](image)

**Figure 13.** The dependence of microhardness on the number of thermal cycles in various states.
The most sensitive characteristic to thermal cycling is the yield strength. In both states of the alloy, it increases with an increase in the number of cycles. UFG states up to TC are characterized by higher values of strength and yield strength due to the contribution of grain boundary hardening. The tensile strength in the CG state increased to 150 cycles, then the values decreased and did not change. This is probably due to the fact that after a certain number of cycles (in CG state, \( n = 150 \)), the material is saturated and no further increase in characteristics is observed. In the UFG state, with an increase in the number of cycles of multiple transformations, the values of the tensile strength
increased until reaching 100 thermal cycles; then, a decrease in the parameter was observed, which can be explained by the heterogeneity of the formed structure.

According to the constructed graphs (Figure 16), the functional characteristics (the magnitude of the reversible deformation and the estimated reactive stress) increased slightly and then remained stable both in the CG and in the UFG states with an increase in thermal cycles.

4. Discussion

The results of this study showed that during multiple martensitic transformations, dislocations were generated and accumulated in the Ti-50.8 at.% Ni alloy. This process occurs intensively in the first 100 cycles, after which a saturation effect is observed and the structure and mechanical characteristics practically do not change. The saturation effect also leads to the stabilization of martensite, which is reflected in changes in the temperatures of martensitic transformations. Thus, the obtained results allow us to state that, as a result of thermal cycling to 100–150 cycles, a change in structural parameters and mechanical and functional characteristics is observed, but a subsequent increase in the number of cycles, in general, results in the stabilization of the alloy in both coarse and ultrafine grains. Thus, it can be said that, for this alloy, 100–150 cycles are sufficient to enhance the mechanical and functional characteristics and form a stable structure.

5. Conclusions

1. As a result of thermal cycling in the Ti-50.8 at.% Ni alloy, an increase in the dislocation density occurs, internal stresses in the CG and UFG states increase, the size of the structural components decreases slightly, which is associated with the formation of dislocation walls and sub-boundaries.

2. When studying the microhardness of titanium nickelide in the CG and UFG states as a result of multiple cycles, the values slightly increase compared to the state before thermal cycling. The first 100 cycles are characterized by a more intense increase in microhardness, then the values stabilize.

3. As a result of mechanical tensile tests for the alloy in both CG and UFG states, the mechanical properties increase slightly, especially the yield strengths and phase yield stresses. Saturation occurs after 150 cycles. Furthermore, in the states under study as a result of repeated martensitic transformations, the functional characteristics—the estimated reactive stress and the length of the phase yield area, which determines the magnitude of the reversible deformation—remain stable.

4. The Ti-50.8 at.% Ni alloy in the UFG state is more attractive for applications, since in this state a higher level of properties is obtained compared to the coarse-grained state. In addition, the UFG state shows greater stability during thermal cycling with a large number of cycles.

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