High-Temperature Plasticity of the $\beta$-phase in Nearly-Equiatomic Nickel-Titanium Alloys

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Plastic behavior of the $\beta$ nickel-titanium alloys with nearly-equiatomic compositions, Ti-49.5, 50.0 and 50.5 mol%Ni, has been studied. The alloys were highly ductile due to dislocation glide at temperatures where the $\beta$-phase was in equilibrium. The ductility became largest at the lowest temperature in the stability range of the $\beta$-phase. The strain-rate vs. flow stress relations of easy-glide deformation in tensile testing and steady-state creep deformation satisfied the power law. Both of the deformations are associated with the dislocation glide which are likely to be controlled by the Peierls mechanism. A yield drop at the onset of plastic flow appeared only in the tensile deformation of Ti-50.5 mol%Ni, but not in Ti-49.5 and 50.0 mol%Ni alloys. Excess nickel atoms would exist as antistructure defects, which might be responsible for the yield point phenomena.

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**Keywords:** nickel-titanium alloy, shape memory alloy, $\beta$-phase, high temperature, creep, yield drop, the Peierls stress, structural vacancy

I. Introduction

Among various shape memory alloys, nickel-titanium alloys have excellent mechanical properties, and thus have been the most important alloys for applications\(^{(10)}\). Martensitic transformation occurs in the $\beta$-phase, which is an intermetallic compound with the CsCl(B2)-type order. It has been known that the alloy is much higher in ductility than the other $\beta$-phase of titanium-based alloys, such as TiFe and TiCo\(^{(19)}\), and nickel-based alloy, NiAI\(^{(9)}\).

Rozner and Waselewski reported the tensile properties of equiatomic NiTi alloy in a wide range of temperature\(^{(6)}\). They observed a variety of stress-strain curves. Different types of the curves can be categorized into the following four types according to the deformation temperature: (i) Below the martensitic transformation temperature, which was around room temperature, the specimen was martensitic, and the shape memory effect was observed. (ii) Between room temperature and 373 K, stress-induced transformation occurred. The stress for the transformation increased with increasing temperature, thus a positive temperature dependence appeared. (iii) Above 373 K plastic deformation due to dislocation glide occurred. Up to 673 K, the yield stress was almost independent of temperature. (iv) Above 673 K, the yield stress decreased with increasing temperature as a normal thermally activated deformation. The highest temperature in their experiment was 973 K; the curve at the temperature where the $\beta$-phase was in equilibrium was limited to this. Except for the study, the mechanical behavior of the equilibrium $\beta$-phase has not been reported to the present authors' knowledge.

It has been known that the martensitic transformation temperature of the alloy is very sensitive to its composition; the increase of 0.1 mol% nickel concentration from the equiatomic composition reduces the Ms temperature by about 10 K\(^{(5)}\). Therefore, severe control of composition is required for the production of the shape memory alloy. The $\beta$-phase exists with a certain the homogeneity range\(^{(6)}\), similarly to the $\beta$-NiAl compound\(^{(6)}\). The range of homogeneity expands in the nickel-rich side and becomes narrower as temperature decreases, and the alloy with the equiatomic composition has an eutectoid point at around 900 K with the reaction, NiTi$\rightarrow$NiTi$_2$ + Ni$_3$Ti\(^{(7)}$. It indicates that the phase is most stable in the stoichiometric composition of the CsCl ordered binary alloys.

In the ordered alloy that has a deviation in the composition from its stoichiometry, structural defects exist; the defects are either antistructure defects, which are substituional atoms of the excess element or constitutional vacancies, which are formed in the sublattice of deficient element\(^{(9)}\). It has been shown in $\beta$-NiAl and FeAl alloys\(^{(9)}\), both of which have the CsCl structure, that structural vacancies exist in aluminum-rich composition, and antistructure defects exist in aluminum-poor composition. The previous authors\(^{(9,9)}\) have concluded that the structural vacancies are responsible for their relatively large hardness at low temperatures.

The type and concentration of the structural defects
are different among various compounds, and some of the evidence was obtained for the β-NiTi alloy. Suzuki and Masumoto have shown that excess nickel atoms are likely to be substitutional defects\(^{12}\). Suzuki has shown that a remarkable quench-hardening behaviour, which had been already known in Ti-55 mol\%Ni\(^{13}\), occurred in the composition range from 52 to 55 mol\%Ni\(^{14}\). The precipitation of NiTi\(_2\) and its intermediate reactions\(^{15}\), which have been examined in detail by Nishida et al.,\(^{15}\), might be responsible for the softening of quench-hardened nickel-rich alloys\(^{14}\).

The present study reports the plastic behaviour of NiTi shape memory alloys at high temperatures where the β-phase is in equilibrium. Therefore, the effect of the precipitation on the mechanical properties can be eliminated. Attention will be paid whether there is any evidence for the existence of structural defects and their interaction in the stress-strain behaviour.

II. Experimental Procedure

Titanium-nickel binary shape memory alloys, Ti-49.5, 50.0 and 50.5 mol\%Ni, were supplied by Furukawa Electric Co. Ltd. The as-received samples have a wire shape with a diameter of 0.9 mm. Surface oxidation layer was removed by a solution of hydrofluoric acid: nitric acid = 1:1 in volume. They were annealed at 1173 K for 4.3 × 10\(^4\) s in a vacuum below 1 × 10\(^{-3}\) Pa. Etching using a solution of hydrofluoric acid:nitric acid:glycerine = 1:1:8 in volume revealed grain boundaries in the sectioned area; the average grain size was 15 μm. Tensile tests were performed with an Instron-type testing machine. The specimen was deformed in a vacuum below 1 × 10\(^{-2}\) Pa. The elongation of the gauge part of the specimen, 30.0 mm in length, was measured from the displacement of the cross head. Creep tests with constant load were performed in a vacuum below 1 × 10\(^{-4}\) Pa. Two grips made by stainless steel were spot-welded on both ends of the specimens. The relative displacement of the grips was measured with a dial gauge. Instantaneous creep strain just after loading could not be measured without uncertainty due to the play of the present setups and the measurement started after a few seconds from the loading.

III. Results

1. Tensile deformation in Instron-type testing machine

Tensile tests were performed on Ti-49.5, 50.0, and 50.5 mol\%Ni alloys at temperatures between 800 and 1150 K. Load vs. cross-head displacement curves were obtained by deformation at a constant cross-head velocity of 1.67 × 10\(^{-3}\) m·s\(^{-1}\), and were reduced to true stress-true strain curves. They are shown in Fig. 1. It is found that plastic yielding in Ti-50.5 mol\%Ni alloy was accompanied by a yield drop; however, the yieldings in Ti-49.5 and 50.0 mol\%Ni alloys were not. It was common to all the alloys that no work-hardening occurred; as the tensile strain increased, the flow stress gradually decreased until fracture occurred. Above 900 K the fracture strain became larger with decreasing the test temperature, as shown in Fig. 2. The strain at 800 K was much smaller than at 900 K, so that a peak appeared around 900 K in each temperature vs. fracture strain relation. The temperatures giving the peaks were different among these compositions. They almost coincided with the lowest temperature limits of the stability of the β-phase in the phase diagram\(^{16}\).

The temperature dependence of the yield stresses is shown in Figs. 3(a) and (b). The yield stresses of Ti-50.0 mol\%Ni alloy in the present study and those obtained by Rozner and Wasilewski\(^{16}\) are plotted by open and closed symbols, respectively, in Fig. 3(a). The stress-strain behaviours observed by the previous authors can be distinguished into the four types as mentioned before. The solid lines in the figures (a) and (b) trace the plots of the present results by the function,

\[
\sigma_y = A / T^\alpha,
\]

where \(\sigma_y\) is the yield stress in MPa; \(T\) is temperature in K; and \(A\) and \(\alpha\) are constants with the values \(A = 1.73 \times 10^{15}\), \(\alpha = 4.51\), in the case of Ti-50.0 mol\%Ni alloy above 900
K. A simple extrapolation of the curve into the temperature region below 900 K, the dotted curve, shows that the yield stress below 900 K is lower than the stress expected from the extrapolation. The yield stresses of Ti–49.5 and 50.0 mol%Ni alloys were almost equal, and also equal to the lower yield points in the Ti–50.5 mol%Ni alloy. The yield drop in the Ti–50.5 mol%Ni alloy was not so sharp; it occurred in a range of a few percent strains. The amount of the drop depended on temperature; it decreased with increasing the test temperature. Whereas, the ratio \( \sigma_{\text{yield}} / \sigma_{\varepsilon} \) was almost constant. It was 1.42 at 900 K, and 1.41 at 1150 K.

The deformation started at a strain-rate of \( 5.6 \times 10^{-4} \) s\(^{-1} \). When specimen is deformed by a constant cross-head velocity, the strain-rate should decrease as the tensile strain increases. Then, the flow stress will change during the deformation owing to its dependence on the strain-rate. To eliminate the effect included in the stress-strain curves of Fig. 1, the curves were transformed by the following procedure: if the volume of specimen is constant during deformation, the strain-rate \( \dot{\varepsilon} \) at a strain \( \varepsilon \) is given by

\[
\dot{\varepsilon} / \dot{\varepsilon}_0 = \exp (-\varepsilon).
\]

Then, if the flow stress \( \sigma \) and the strain-rate obey the \( m \)-th order power law,

\[
\dot{\varepsilon} = K \sigma^m,
\]

where \( K \) is a constant, the flow stress can be related to the strain as

\[
\sigma / \sigma_0 = \exp (-\varepsilon / m),
\]

where \( \sigma_0 \) is the flow stress at \( \varepsilon = 0 \). If we neglect elastic strain, the tensile strain \( \varepsilon \) is equal to plastic strain. Then, we take yield stress \( \sigma_y \) for the initial stress \( \sigma_0 \).

**Figure 4** shows the change in \( \ln(\sigma / \sigma_y) \) as a function of plastic strain \( \varepsilon \) in Ti–50.0 mol%Ni alloy. Plateau regions appeared in the relations above 900 K. In the plateau the flow stress slightly decreased with increasing strain, and the power law equation was satisfied. The stress-exponent \( m \) was determined from the slope of the plateau to be \( 5.7 \pm 0.4 \) in Ti–49.5 mol%Ni, \( 5.7 \pm 0.3 \) in Ti–50.0 mol%Ni and \( 5.5 \pm 0.3 \) in Ti–50.5 mol%Ni alloys. Beyond the extent of the plateau the exponent \( m \) decreased with increasing tensile strain. The extent of the power law deformation depended on temperature; the dependence was similar to that of the fracture strain shown in Fig. 2.

**Figure 5** shows the strain-rate vs. flow stress relations obtained by the tensile testings at 1100 K. They were measured by changing the cross-head velocity abruptly during the tensile testing. Linear relations were obtained between \( \ln \sigma \) and \( \ln \dot{\varepsilon} \) at the stress above 20 MPa. The stress-exponent of the power law is \( 5.94, 5.88 \) and \( 5.78 \) in Ti–49.5, 50.0 and 50.5 mol%Ni alloys, respectively. The stress-exponent in low strain-rate region is nearly unity. The fact suggests\(^{(10)} \) that the deformation mechanism in this region might be diffusion creep.
2. Creep test at constant load

Creep tests were made by applying constant load, where the stress at the beginning of the creep was 17, 25 or 33 MPa. Judging from the result of the strain-rate changing test, we can expect that dislocation creep will occur at 25 MPa and 33 MPa, and diffusion creep at 17 MPa. Typical creep curves are shown in Figs. 6(a) and (b); the initial stress $\sigma_0=33$ MPa and the creep temperature $T=1050$ K in (a); $\sigma_0=17$ MPa, $T=1250$ K in (b). In every creep temperature and initial applied stress, the creep-rate of Ti-50.5 mol%Ni alloy was largest and the rate of Ti-49.5 mol%Ni alloy smallest. The curves in Fig. 6(a) appears to consist of three stages corresponding to transient, steady-state and accelerating creep, which are denoted by Stage I, II and III, respectively. In contrast, the curves in Fig. 6(b) are parabolic. This type of curve was typically observed in the creep of $\sigma_0=17$ MPa.

In the case of creep under constant load, the stress should increase with increasing the strain, so that steady-state deformation with a constant creep-rate does not appear. The three stages in Fig. 6(a) were distinguished by the following procedures: Under a constant load, the creep stress and strain satisfy the relation
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\[ \frac{\sigma}{\sigma_0} = \exp \varepsilon. \]  \hspace{1cm} (5)

If the power law of stress-exponent $n$ holds between the creep-rate and stress, the creep-rate is given by

\[ \dot{\varepsilon} / \dot{\varepsilon}_0 = \exp (n \varepsilon). \]  \hspace{1cm} (6)

In Figs. 7(a) and (b) is plotted $\ln(\dot{\varepsilon} / \dot{\varepsilon}_0)$ as a function of $\varepsilon$. Figure 7(a) shows the plots of the creep of $\sigma_0 = 25$ MPa in Ti-50.0 mol\%Ni alloys, and (b) those of $\sigma_0 = 33$ MPa and $T = 1000$ K in Ti-49.5, 50.0, 50.5 mol\%Ni alloys. After revealing a transient creep, a linear portion appeared in each plot. In this region, a power law deformation could be assumed. In this paper, secondary creep (Stage II) is designated as power law creep. The creep-rate showed the minimum at the onset of the linear portion, and is pointed out by an upward arrow. The extent of the power law region depended on the creep temperature and the composition; it is similar to the change in the fracture strain in tensile testing.

The linear portion did not appear in the plots of the creep curves of $\sigma_0 = 17$ MPa. As mentioned above, diffusion creep would occur at the beginning of the creep; however, as the creep strain increased, the increase of applied stress due to the reduction of cross section could induce dislocation-gliding. For example, eq. (5) predicts that the stress arises from 17 to 25 MPa by the creep strain of 0.4. Therefore, conventional constant load test is not appropriate for the observation of diffusion creep at high initial strain, and we excluded the analysis of diffusion creep from the present report.

The stress-exponent of the power law deformation was measured from the slope of each plot, and plotted as a function of the creep temperature in Fig. 8. They were smaller than the exponents in the tensile tests. The exponents of Ti-50.0 and 50.5 mol\%Ni alloys did not depend on the creep temperature in the $\beta$-phase region; they were scattered within the range of $\pm 0.2$ around 2.5. The exponent of Ti-49.5 mol\%Ni alloy deviated further from the value with decreasing creep temperature below 1050 K. Although the power law deformation indeed appeared, the linear portion in the $\ln(\dot{\varepsilon} / \dot{\varepsilon}_0)$ vs. $\varepsilon$ plot was much narrower than that above 1050 K.

Activation energy of the power law creep was calculated from the Dorn-type equation\(^\text{(17)}\); the energy $Q_\varepsilon$ in J mol$^{-1}$ was defined through the minimum creep-rate $\dot{\varepsilon}_\text{min}$.
Fig. 9 $\dot{\varepsilon} - T$ vs. $1/T$ relations on minimum creep rate $\dot{\varepsilon}_{\text{min}}$ (a), and the initial creep rate $\dot{\varepsilon}_0$ (b).

Table 1 Activation enthalpy of creep.

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Stress (MPa)</th>
<th>Activation enthalpy (kJ·mol$^{-1}$)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>Initial rate</td>
</tr>
<tr>
<td>Ti-49.5 at%Ni</td>
<td>33</td>
<td>169</td>
</tr>
<tr>
<td>Ti-50.0 at%Ni</td>
<td>33</td>
<td>151</td>
</tr>
<tr>
<td>Ti-50.5 at%Ni</td>
<td>25</td>
<td>154</td>
</tr>
</tbody>
</table>

in s$^{-1}$ in the form

$$\dot{\varepsilon}_{\text{min}} = A \mu b / kT \cdot \exp \left(-\frac{Q_c}{kT}\right),$$

where $\mu$ is the shear modulus; $b$, the Burgers vector of dislocation; $k$, the Bolzmann constant; $A$, constant. Since the modulus $\mu$ depends on temperature, we should take into account for the change in $\mu$. However, no precise form of the function has been known, so that we neglected the contribution to the calculation of $Q_c$ as an approximate value. Activation energy was also calculated for the initial creep-rate $\dot{\varepsilon}_0$ as indicated by arrows in Fig. 7(a) and (b), for example. Figures 9(a) and (b) show the changes in $\dot{\varepsilon}_{\text{min}} - T$ and $\dot{\varepsilon}_0 - T$ with $1/T$, respectively. The difference between $\dot{\varepsilon}_{\text{min}}$ and $\dot{\varepsilon}_0$ became smaller with increasing the creep temperature. Namely, as the temperature is higher, the transient creep comes to appear lesser. It was confirmed in the creep of Ti-50.5 mol%Ni alloy at $\dot{\varepsilon}_0 = 33$ MPa and 1050 K that no transient creep appeared, and power law creep occurred from the beginning. The energies obtained from the Arrhenius plots are listed in Table 1; the values for the initial creep-rate $\dot{\varepsilon}_0$ were close to the activation energy of diffusion in the $\beta$-NiTi alloy, 138 KJmol$^{-1}$ (1.43 eV), determined by Bastin and Rieck$^{188}$. On the other hand, although the Dorn-type equation assumes the activation energy for that of diffusion, the energies of the power law creep were much higher.

**IV. Discussion**

1. Composition-dependence of plastic deformation behaviour

The differences in the plastic deformation behaviour among the alloys with the compositions, Ti-49.5, 50.0 and 50.5 mol%Ni, were confirmed in the following aspects; (i) fracture strain, (ii) upper bound strain of power law creep (secondary creep), (iii) creep-rate, (iv) yield point phenomena. Each of the former three terms, (i), (ii), (iii), can be expressed as an increasing function of nickel content. However, the term (iv) was observed only in the alloy with an excess nickel content. Thus the origin seems to be different from those for the former three.

In the most part of the deformation in tensile and creep, the power law was satisfied. The stress-exponents for both deformations appeared to be different among the alloys, respectively; however, the difference was small. The creep-rate of Ti-49.5 mol%Ni was always lower than the rates of the others, although the stress-exponent was slightly larger. The fact implies that the difference in the creep-rate did not come from the difference in the exponents, but from the difference in the constant $K$ in eq. (3). The magnitude of $K$ might be determined from some intrinsic factors of dislocations' mobility, such as stacking-fault energy and the Peierls stress. The experimental results in the present study were not enough for the further discussion of the composition-dependence of plasticity on account of the factors.

On the yield point phenomena in intermetallic compounds, Besag and Smallman discussed all the possible mechanisms$^{19}$. Among the mechanisms dislocation-locking due to the formation of Cottrell atmosphere by impurities, carbon or oxygen atoms, can be excluded in the present case, since the yield drop occurred only in Ti-50.5 mol%Ni alloy and these impurities would not be present in the matrix due to the formation of stable precipitates, TiC and Ti$_4$Ni$_3$O, during solidification$^7$.

The observed yield drop was not sharp, and had a width of about 3 to 4% strain. Moreover, the stress-drop strongly depended on temperature. These characteristics are similar to that of Ni$_3$Al$^{225}$ alloy, which has been concluded to be the Johnston-Gilman type$^{226}$, but not the Cottrell-Petch type. The former type assumes that dislocations glide in a viscous manner, and that yielding occurs when multiplication of dislocation leads the dislocation density to exceed a critical magnitude. Dragging of the atmosphere of point defects around dislocations is a probable mechanism, and high concentration of structural defects in the Ti-50.5 mol%Ni alloy might be responsible for it.

When the average velocity of dislocations depends on stress by the $m$-th order power law, the ratio of dislocation densities at upper and lower yield points, $\rho_u$ and $\rho_l$, can be related to the ratio of the upper yield stress $\sigma_u$ to
the lower yield stress \( \sigma_1 \), as
\[
\rho_1 / \rho_0 = (\sigma_1 / \sigma)^m. \tag{8}
\]
Using the value of \( m \) measured in the tensile test at 1100 K; \( m = 5.8 \), the ratio \( \rho_1 / \rho_0 \) is 7.4. Since the ratio \( \sigma_1 / \sigma \) was almost constant for the test temperature above 900 K, the density-ratio before and after the yielding was also constant. The multiplication of such amount, not more than the factor of 10, can be easily attained by the onset of plastic deformation.

2. Power law deformation due to dislocation glide

In the case of power law creep in metals and alloys, the stress-exponent is typically 5 in pure metals, and 3 in solid solutions\(^{17,22,33}\). The exponent of 2.5 in the \( \beta \)-NiTi alloys suggests that the creep mechanism is close to the latter class, where viscous motion of dislocations is the rate controlling process.

Prior to secondary creep, transient creep was observed with a strain up to 0.1. The creep-rate decreased in this stage. The activation energy of the transient creep indicated that a diffusion-controlled process occurred; one of the probable processes is micro creep, where the motion of pre-existing dislocations are controlled by the rate of their climb and they are exhausted during the stage. The theory of Weertman gives the explanation of parabolic creep curves by the microcreep\(^{24}\).

The stress-exponent of the power law in the tensile deformation differed from the exponent in the power law creep. The mechanism of the easy glide deformation is not necessary to be identical to that of steady-state creep, where the rate of work-hardening and dynamic recovery are balanced. Rather, the fact that the tensile stress did not strongly depend on the amount of plastic deformation would be a result of no work-hardening during the deformation. It might be be due to planar configuration of dislocations and the cross slip of screw dislocations with small stress. Then, the strain-rate of the easy glide would be governed by the average velocity of dislocations \( v \) through the Orowan's equation, \( \dot{\varepsilon} = p b v \), where \( p \) is the dislocation density, and the power law might originate from the stress dependence of the velocity, \( v \propto \sigma^m \).

To compare with the activation energy of power law creep, the activation enthalpy of the easy glide deformation \( Q_e \) was estimated from the equation proposed by Mott\(^{25}\),
\[
\dot{\varepsilon} = A \cdot \exp \left( - \frac{Q_e}{kT} \right), \tag{9}
\]
where \( A = A(\sigma) \) is a function of stress. When the power law of \( m \)-th order holds, the enthalpy is given as
\[
Q_e = -m k T \frac{d(\sigma d/\sigma)}{dT}, \tag{10}
\]
where the relation, \( d \ln \dot{\varepsilon} /dT = -d(\ln \dot{\varepsilon} /d\sigma) \cdot (\sigma d/\sigma) \), is used. If we assume that eq. (9) is an Arrhenius-type function such that \( Q_e / kT \) does not apparently depend on \( T \), the flow stress \( \sigma \) can be expressed in the form of eq. (1); \( \alpha = 4.39 \) in Ti-49.5 mol%Ni; 4.51 in Ti-50.0 mol%Ni; 4.65 in Ti-50.5 mol%Ni. Then, the enthalpy can be expressed as

\[
Q_e = m k T. \tag{11}
\]
Using the exponents \( m \)'s at 1100 K, the enthalpy is calculated to be 234 kJ mol\(^{-1}\) (2.42 eV) in Ti-49.5 mol%Ni; 245 kJ mol\(^{-1}\) (2.54 eV) in Ti-50.0 mol%Ni; 248 kJ mol\(^{-1}\) (2.54 eV) in Ti-50.5 mol%Ni alloys. These magnitudes are close to the activation enthalpies of the secondary creep.

From the above assumptions, apparent activation volume \( V^* = kT(\partial \ln \dot{\varepsilon} / \partial \sigma) \) is deduced to
\[
V^* = m k T / \sigma. \tag{12}
\]
Taking the Burgers vector of the superdislocation in the CsCl-ordered lattice as \( b = a_0 [111] \); \( a_0 = 0.301 \) nm the lattice parameter of the \( \beta \)-NiTi\(^{30}\), the apparent activation volume of the tensile deformation and creep at \( T = 1100 \) K, \( \sigma_0 = 33 \) MPa is \( 21 b^2 \) and \( 9 b^2 \), respectively.

The classification of Conrad\(^{26}\) implies that the deformation mechanism giving activation volume of the order of \( 10^2 b^2 \) or lesser is either cross-slip or the Peierls mechanism. The strong temperature dependence of the flow stress, as shown in Fig. 3, suggests that the latter mechanism is probable. In some bcc alloys\(^{27}\) and intermetallic compounds, such as Ni\(_3\)Al, Ni\(_3\)Ti\(^{21}\) and TiSi\(_2\)\(^{28}\), it has been concluded that the flow stress is governed by the Peierls mechanism even at high temperatures. According to the theory of Weertman\(^{20}\) referred above, the process controlled by the formation of kink-pairs over the Peierls potential has the stress-exponent of 2.5. Therefore, the creep of the \( \beta \)-NiTi alloys can be well described by his lattice friction model.

As the strain went beyond the bound of the power law deformation, the exponent decreased in the tensile deformation, and increased in the creep test. The difference can be understood from the difference in the modes of the two deformations. In the case of tensile testing, the strain-rate should decrease with increasing tensile strain, whereas, in the creep test by constant load, applied stress should increase. Therefore, upper bound strain of the power law is due to the transition of the deformation mechanism into diffusion creep in tensile, and into high stress-exponent deformation\(^{17,22}\) in creep.

V. Concluding Remarks

(1) The stress-strain curve in tensile deformation showed no work-hardening, and the flow stress satisfied power law with the stress exponent of about 5.8. The yield stresses in the \( \beta \)-phase of Ti-49.5 and 50.0 mol%Ni alloys and the lower yield points in Ti-50.5 mol%Ni alloy were almost equal. Whereas, fracture strain depended on the composition; it was larger in the alloy with higher nickel content. The strain became larger as the test temperature was lower within the stable region of the \( \beta \)-phase.

(2) Power law creep (Secondary creep) with the stress-exponent of 2.5 was observed in the constant load creep test when the initial stress was above 20 MPa. The activation energy of transient creep was equal to the ac-
tivation energy of diffusion. The activation enthalpy of the power law creep, 240 kJmol\(^{-1}\), might be that of the deformation controlled by the Peierls mechanism.

(3) Yield drop occurred in the tensile deformation of Ti-50.5 mol% Ni alloy, but not in the deformation of Ti-49.5 and 50.0 mol% Ni alloys. Structural defects in the alloy having excess nickel content than equiatomic composition might be responsible for it.

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