Effect of swaging on Young's modulus of β Ti-33.6Nb-4Sn alloy

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ABSTRACT

The effect of swaging on the Young's modulus of β Ti-33.6Nb-4Sn rods was investigated by X-ray diffraction, thermography, microstructural observations, deformation simulator analysis and cyclic tensile deformation. Stress-induced α'' martensite was stabilized by swaging, dependent on the diameter reduction rate during swaging. Thermography and deformation simulator analysis revealed that swaged rods were adiabatically heated, and consequently, stress-induced α'' underwent reverse transformation. Young's modulus, which was measured by the slope of the initial portion of the stress-strain curve, decreased from 56 GPa in the hot-forged/quenched rod to 44 GPa in the rapidly swaged rod with a high reduction rate and to 45 GPa in the gradually swaged rod with a low reduction rate. The tangent modulus, which was measured by the slope of the tangent to any point on the stress-strain curve, decreased with strain even in the linear range of the stress-strain curve of the hot-forged/quenched rod and the rapidly swaged rod, while the tangent modulus remained unchanged for the gradually swaged rod. It was found that Young's moduli in swaged β Ti-33.6Nb-4Sn rods were affected by stabilized α'' martensite. Low Young's modulus of 45 GPa and high strength over 800 MPa were obtained when the reverse transformation by adiabatic heating was suppressed and the stress-induced α'' was sufficiently stabilized by gradual swaging to a 75% reduction in cross section area.

Keywords: beta titanium alloy, swaging, Young's modulus, strength, stress-induced martensite, adiabatic heating

1. Introduction

Extensive studies on β Ti alloys with low Young's modulus have brought out many alloys for biomedical applications to avoid stress shielding arising from a large difference in Young's moduli between $\alpha+\beta$ Ti alloys such as Ti-6Al-4V (100-110 GPa) and cortical bone (10-30 GPa). Among them several studies on the elastic moduli of single crystals are helpful in developing advanced medical devices. Tane et al. (Tane et al., 2008) revealed the elastic anisotropy of $E_{100} < E_{110} < E_{111}$ in β Ti-Nb-Ta-Zr, where E_{100} , E_{110} and E_{111} were Young's moduli in the $\langle 100 \rangle$, $\langle 110 \rangle$ and $\langle 111 \rangle$ directions, respectively. Similar elastic anisotropy was reported in β Ti-Nb-Al (Inamura et al., 2005), β Ti-24Nb-4Zr-8Sn (Zhang et al., 2011) and β Ti-Mo-Zr-Al (Lee et al., 2012). Lee et al. summarized the data on the elastic moduli of β Ti alloy single crystals, and claimed that a low Young's modulus was strongly correlated to low electron to atom ratio, e/a , ranging from 4.1 to 4.8. If this is the case, $\langle 100 \rangle$ -oriented β Ti alloy single crystals with a low e/a ratio as a candidate material can be used to fabricate medical devices. However, single crystals are costly and have insufficient strength for structural implants.

Recently Jung et al. (Jung et al., 2013) investigated the mechanical properties of severely swaged β Ti-25Nb-11Sn rods and found that the rods consisting of a β single phase had low Young's modulus of 53 GPa and high strength over 800 MPa. Their alloy possessed a small e/a of 4.16 and a strong $\langle 110 \rangle$ fiber texture along the swaging direction. According to Zhang et al. (2011), the elastic moduli of Ti-24Nb-4Zr-8Sn with an e/a of 4.15 are $E_{100} = 27.1$ GPa, $E_{110} = 56.3$ GPa and $E_{111} = 88.1$ GPa. It is interesting to note that Young's modulus of the swaged β Ti-25Nb-11Sn rods with $\langle 110 \rangle$ fiber texture is close to that of $[110]$ oriented Ti-24Nb-4Zr-8Sn single crystal. Although the utilization of $\langle 100 \rangle$ deformation texture is promising for the application of β Ti alloy polycrystals to medical devices with a low Young's modulus and high strength, most of the deformation textures observed in β Ti alloys are not $\langle 100 \rangle$, but are $\langle 110 \rangle$ along the rolling or swaging direction.

Matsumoto et al. (2005) and Jung et al. (2012) investigated the effect of stress-induced α'' martensitic transformation on the Young's modulus using β Ti-33.6%Nb-4%Sn with an e/a of 4.21. They found that cold rolling leads to a low Young's modulus of less than 50 GPa and interpreted that the low Young's modulus is associated with $\langle 110 \rangle_{\beta}$ and $\langle 010 \rangle_{\alpha''}$ texture formation along the rolling direction. However, because there are no available data on elastic moduli of α'' martensite, the cause of a low modulus in the rolled β Ti-33.6%Nb-4%Sn is not completely understood.

There are two possible methods to further decrease the Young's modulus of β TiNbSn alloys. One method is to modify the alloy composition of a single β phase to approach a small e/a value of approximately 4.1, according to Lee et al. (2012) and the other is to use stress-induced α'' martensitic transformation. This study focused on the latter method to examine the Young's modulus of severely swaged β Ti-33.6%Nb-4%Sn and discussed a mechanism of the low Young's modulus in comparison with the swaged Ti-25Nb-11Sn in which α'' martensite was not stress-induced (Jung et al., 2013). It should be noted that when β Ti-33.6%Nb-4%Sn has M_s (martensitic transformation start) temperature just below room temperature, stress-induced α'' martensite readily undergoes reverse transformation on heating to relatively low temperatures. This indicates that the adiabatic temperature increase occurring during swaging may enhance the reverse transformation. Therefore, Young's modulus was investigated at two diameter reduction rates leading to different adiabatic heating in swaging.

2. Experimental procedure

Five Ti-Nb-Sn alloy ingots weighing 12 kg per ingot with the nominal composition of Ti-33.6mass%Nb-4mass%Sn were prepared by high frequency induction melting in vacuum using a water-cooled copper hearth, followed by vacuum arc remelting to produce a 60 kg ingot with a diameter of 200 mm ϕ . The ingot was hot-forged at starting temperature of 1370 K to a diameter of 38 mm ϕ and water quenched. This procedure will be referred to "hot-forged/quenched" hereafter. Surface oxide layers were mechanically removed by scalping to a 32 mm ϕ rod for swaging experiments. The chemical composition after scalping was Ti-33.8mass%Nb-3.96mass%Sn with impurities of 0.008%C, 0.058%O, 0.008%N and 0.02%Fe. The interstitial impurities were measured using the combustion infrared detection for C (CS444LS, LECO corp., St. Joseph, USA), the inert gas fusion infrared detection for O (TC600, LECO corp.) and the inert gas fusion thermal conductivity detection for N (TC600, LECO corp.). Hereafter, this alloy will be referred to as Ti-33.6Nb-4Sn. Swaging from 32 mm ϕ to 15.9 mm ϕ corresponding to a 75% reduction in cross section area was performed gradually and rapidly to examine the effect of the swaging rate on the adiabatic temperature increase. During gradual swaging, 13 types of swaging dies with different sizes for 30.0, 28.8, 27.3, 25.9, 24.5, 23.2, 22.0, 20.8, 19.8, 18.7, 17.7, 16.8 and 15.9 mm ϕ rods were used to gradually decrease the rod diameter. During rapid swaging, 2 types of swaging dies for 20.0 and 15.9 mm ϕ rods were employed to rapidly decrease the rod diameter. It took approximately 5 min to replace the dies in each swaging pass of the gradual and rapid swaging.

The rod temperature just before and immediately after swaging was measured by

thermography (i40, FLIR Systems, Inc., Wilsonville, USA). To ensure measurement accuracy, one end of a Ti-33.6Nb-4Sn rod with 9 mm in diameter and 100 mm in length was locally heated to 573, 623 and 673 K. The generated temperature gradation was simultaneously measured by a thermographic camera and thermo-couples, and the data measured by thermography were calibrated by temperature measurements using thermo-couples. The transformation temperatures of $\beta \rightarrow \alpha''$ and $\alpha'' \rightarrow \beta$ were measured by a differential scanning calorimeter (DSC 204F1 Phoenix, Netzsch Japan) at heating and cooling rates of 10 and 30 K/min. The microstructures of the alloy were observed by optical microscopy (OM). Constituent phases and textures before and after swaging were measured by X-ray diffraction (XRD) with Cu-K $_{\alpha}$ radiation. Tensile samples with a gage portion of 20 mm in length and 3 mm in diameter were prepared from the hot-forged/quenched and the swaged rods. Tensile tests to failure at an initial strain rate of $8 \times 10^{-4} \text{ s}^{-1}$ were carried out at 295K with a tensile testing machine (Autograph AG-IS, Shimadzu Corp., Kyoto, Japan) equipped with a video extensometer. Cyclic tensile tests at a strain rate of $4 \times 10^{-4} \text{ s}^{-1}$ were performed at 295K with a fatigue testing machine (Servopulser EHF, Shimadzu Corp.). To ensure the accuracy in measuring the modulus by stress-strain curves, tension/compression jigs with high rigidity were equipped to the testing machine, and tensile axis alignment was carefully adjusted so that the measured strain was not affected by any insufficient rigidity of the machine itself as well as its mechanical play. The strain was precisely measured by a clip-on extensometer. The Young's moduli were calculated from the linear slope of the initial stage of a stress-strain curve. The tangent moduli were measured by the slope of the tangent to any point on the stress-strain curve. The Young's modulus corresponds to the constant tangent modulus at the initial stage of deformation. A deformation simulator (ThermecMaster Z, Fuji Electronic Industrial Co. Ltd., Tsurugasaki, Japan) was used to examine the effect of strain rate on adiabatic heating and microstructure and to discuss the moduli of Ti-33.6Nb-4Sn swaged at different rates. For this experiment two types of cylindrical samples with dimensions of 8 mm in diameter and 12 mm in height and 6 mm in diameter and 9 mm in height were prepared from recrystallized rods heat treated at 1073K after 75% gradual swaging. Prior to compression tests a thermo-couple with 0.5 mm ϕ was inserted in a hole with 0.7 mm ϕ which was horizontally drilled to a 2 mm depth at 1/2 the height of the cylindrical samples.

3. Results and discussion

An ingot prepared by high frequency induction melting followed by vacuum arc remelting was hot-forged at 1373 K and water quenched from approximately 950 K (above the β transus temperature of 870 K) immediately after hot-forging to obtain fine

β grains without β decomposition. Fig. 1 shows OM micrographs of the hot-forged/quenched rod in the (a)transverse and (b)longitudinal sections, where the forging direction is normal to the micrograph in (a) and horizontal in (b). Mixed grain structures consisting of fine, equiaxed (less than 20 μm) grains and coarse, elongated (larger than 100 μm) grains were seen in both sections. Most grain boundaries of the coarse grains elongating along the forging direction were not smooth, but irregular, suggesting that dynamic recrystallization occurred during hot-forging. Fig. 2 shows the results of the DSC measurements of M_s and A_f ($\alpha'' \rightarrow \beta$ transformation finish temperature) of the hot-forged/quenched rod. A DSC sample prepared from the hot-forged/quenched rod was heated to 423 K and cooled to 123 K at 30 K/min(Fig. 2(a)) and then heated to 423 K at 10 K/min (Fig. 2(b)). Although the exothermic reaction indicating $\beta \rightarrow \alpha''$ transformation in Fig. 2(a) was weak, it was found that M_s existed at around room temperature. Actually, M_s was estimated to be just below room temperature, because the OM (Fig. 1) and XRD analyses (described later) showed no evidence of the existence of α'' . In addition to the weak exothermic reaction, the reaction can be observed over a broad temperature range. This result is due to the large distribution of grain size seen in Fig. 1(b), because M_s is dependent on the grain size (Yang and Bhadeshia, 2009). Corresponding to the exothermic reaction, the endothermic reaction indicating $\alpha'' \rightarrow \beta$ reverse transformation can be observed in a broader temperature range, Fig. 2 (b).

The hot-forged/quenched rod was both gradually and rapidly swaged at room temperature. Thermography of the swaged rods revealed that the surface temperature increased by swaging with an increasing reduction in the cross section area(Fig. 3).In gradual swaging the temperature gradually increased and saturated with each swaging pass, whereas in rapid swaging, the temperature increased rapidly over 500 K during the first pass and slightly decreased at the second pass.

Fig. 4 shows the OM micrographs of (a, b) a 75%gradually swaged rod and (c, d)a 75% rapidly swaged rod in the (a, c)transverse and (b, d)longitudinal sections. No apparent difference in the micrographs can be observed between gradual and rapid swaging. Common characteristic features were extremely curled grains in the transverse sections and fiber structure in the longitudinal sections. Similarly curled grains were first found in drawn and swaged tungsten and iron(Peck and Thomas, 1961).Since then this microstructure was reported in many swaged and extruded bcc alloys such as Nb-Tialloys (Naka et al., 1995, Naka and Khan, 2002, Alla mesh et al., 2002), Gum metal(Saito et al., 2003, Furuta et al., 2005, Guo et al., 2008) and Ti-Nb-Zr-Sn (Hao et al., 2012). Naka et al. first termed this microstructure as the VGS (van Gogh's Sky)microstructure. The VGS microstructure formation was consistently explained in terms of plane-strain deformation resulting from $\langle 110 \rangle_\beta$ fiber texture and grain

boundary compatibility (Hosford, 1964). Recently, computer simulations have predicted the VGS microstructure formation sequence by considering a $\langle 110 \rangle_\beta$ fiber texture (Očenášek et al., 2007, Ripoll and Očenášek, 2009). To understand the VGS micrographic features XRD profiles were taken from the transverse sections of the rapidly and the gradually swaged rods. The results obtained are shown in Fig. 5, where an XRD profile of the hot-forged/quenched rod is included for reference. The hot-forged/quenched rod consisted of a bcc (β) single phase with a strong $(110)_\beta$ peak (Fig. 5(a)), and the swaged rods were composed of β with a strong $(110)_\beta$ peak and α'' martensite with $(020)_{\alpha''}$ peak (Fig. 5(b)(c)). Fig. 6 shows the pole figures corresponding to Fig. 5(b)(c), indicating the preferred development of $\langle 110 \rangle_\beta$ and $\langle 010 \rangle_{\alpha''}$ textures. Although the intensity of α'' was much stronger in gradual swaging (Fig. 5(c)) than in rapid swaging (Fig. 5(b)) the characteristic features of the VGS structure were similar to each other (Fig. 4 (a, c)). This result suggests that α'' is very ductile, similar to β during swaging, and does not interrupt the VGS formation. Therefore, it is considered that the VGS microstructure in the swaged Ti-33.6Nb-4Sn (Fig. 4(a)(c)) originates from the development of a $\langle 110 \rangle_\beta$ fiber texture.

Fig. 7 shows stress-strain curves of the hot-forged/quenched, the rapidly swaged and the gradually swaged Ti-33.6Nb-4Sn rods. Deformation behavior is quite different from each other, especially during the initial deformation stage. Double yielding points, which can be related to the stress-induced α'' martensitic transformation for the first yielding and dislocation slip for the second yielding, are observed in the hot-forged/quenched and the rapidly swaged rods. The hot-forged/quenched rod exhibits a lower tensile strength and a larger elongation than the rapidly swaged rod. In contrast, high tensile strength is observed without double yielding in the gradually swaged rod. The initial stage of the stress-strain curves was examined by cyclic deformation to clarify the effect of stress-induced α'' on Young's modulus. The cyclic deformation tests were carried out to 10 cycles, and representative stress-strain curves will be shown at cycles 1, 2, 5 and 10. Fig. 8 shows the cyclic stress-strain curves on the loading/unloading of the hot-forged/quenched Ti-33.6Nb-4Sn rods, where the maximum tensile loading was fixed at a stress between the first and the second yielding. Super elasticity with a small residual strain (ϵ_r) and considerable stress hysteresis were clearly observed during the first cycle. With increasing cyclic number a critical stress (σ_m) for inducing α'' transformation decreased, and the stress hysteresis reduced. ϵ_r increased monotonically with cyclic deformation, although the hysteresis loop was nearly completely closed on loading and unloading. This implies that the stress-induced α'' transformation is accompanied with slip deformation. As a result, an internal stress field will be generated by these dislocations. Stress-induced $\beta \rightarrow \alpha''$ transformation in the following cycle will be assisted by this stress field formed in the prior cycle, thereby leading to the

decrease in σ_m . Such a decrease in σ_m has been observed in many super elastic alloys (Miyazaki et al., 1986, Strnadel et al., 1995, Tahara et al., 2009, Ma et al., 2010, Al-Zain et al, 2012).

Fig. 9 shows an initial stage of the stress-strain curves on the loading of the hot-forged/quenched rods at cycles 1 and 2. Hooke's law seems to hold to a strain indicated by an arrow. If the slope of the initial dotted line in Fig. 9 is assumed to correspond to elastic deformation, the Young's modulus at cycle 1 is measured to be 56 GPa. Fig. 10 shows the tangent modulus measured by the slope of the tangent to any point on the stress-strain curve in Fig. 9. It should be noted that the tangent modulus of cycle 1 is constant at 55 GPa in the strain range from 0 to 0.08% and gradually decreases with increasing strain even in the linear range (0 to 0.2% strain) of the stress-strain curve in Fig. 9. A decrease in the tangent modulus in the linear range of the stress-strain curve was observed more significantly in cycle 2, where a constant modulus disappeared. σ_m , which was determined from an intersecting point of two dotted straight lines, decreased abruptly at the cycle 2. Fig. 11 shows cyclic stress-strain curves on the loading/unloading of the rapidly swaged rods, where maximum tensile loading was fixed at a stress between the first and the second yielding. ϵ_r during the first cycle was larger than that of the hot-forged/quenched alloy (Fig. 8(a)), and increased slightly from 0.7% strain at cycle 1 to 1.1% strain at cycle 10. Stress hysteresis decreased with an increasing number of cycles (Fig. 11(b)-(d)). Fig. 12 shows an initial stage of the stress-strain curves on the loading of the rapidly swaged rod at cycles 1 and 2. The stress-strain curve at cycle 1 showed an initial linear range, where the Young's modulus was measured to be 44 GPa. In contrast, the stress-strain curve at cycle 2 exhibited parabolic behavior even at the initial stage of deformation and it was difficult to determine σ_m by the conventional intersection method. Fig. 13 shows the tangent modulus of the rapidly swaged rod at cycles 1 and 2. The tangent modulus at cycle 1 decreased significantly from 44 GPa with an increasing strain even in the linear range of the stress-strain curve. Therefore, the Young's modulus of 44 GPa measured from the stress-strain curve at cycle 1 in Fig. 12 is apparent, strictly speaking, because a constant tangent modulus cannot be seen. Low modulus was observed at cycle 2 at the initial stage of deformation. Fig. 14 shows cyclic stress-strain curves on the loading/unloading of the gradually swaged alloy, where maximum tensile loading was fixed at a stress similar to the rapidly swaged rod. Both σ_m and stress hysteresis almost disappeared throughout the deformation cycling, and the Young's modulus at cycle 1 was measured at 45 GPa. The tangent modulus on loading of the gradually swaged alloy was almost unchanged and approximately 45 GPa in a strain range of 0~1% (Fig. 15). Fig. 16 summarizes the ϵ_r of the hot-forged/quenched, the rapidly swaged and the gradually swaged rods as a function of deformation cycle. Clearly, ϵ_r exhibited low values in the

swaged rods, especially for the gradually swaged rod. Correspondingly, the stress hysteresis in gradual swaging was very small and became undetectable at cycle 10 (Fig. 14). Therefore, it is evident that the additional formation of stress-induced α'' is negligible on loading and unloading. Though α'' is slightly stress-induced, its formation will be limited to the first few cycles, as evidenced from the stress-strain curves in Fig. 14. In other words, α'' is already stabilized to the highest degree through gradual swaging, and a newly induced α'' is negligible in the subsequent cyclic deformation.

The adiabatic temperature increase during swaging was more remarkable in rapid swaging than in gradual swaging. Temperature over 500 K was reached during the rapid swaging, as shown in Fig. 3. The thermal stability of α'' formed by swaging was investigated by DSC, using the gradually swaged rod. The result obtained is shown in Fig. 17. Evidently, an endothermic reaction indicating $\alpha'' \rightarrow \beta$ transformation started at approximately 350 K and finished at 550 K, which was in fairly good agreement with previous studies in which the thermal stability of α'' formed by sheet rolling was investigated by DSC (Matsumoto et al., 2005) and by XRD in-situ observation during heating (Hanada, 2008). Therefore, it is reasonably explained that α'' induced by rapid swaging undergoes reverse transformation by adiabatic heating. This explanation is consistent with the XRD results that α'' reflection is weaker in the rapidly swaged rod than in the gradually swaged rod (Figs. 5 and 6). If this is correct, a large variety of internal stress fields formed in the α'' transformation will remain after rapid swaging and they will assist stress-induced α'' transformation in the subsequent deformation. Additionally, the interfaces of the α'' plates retained after rapid swaging can move easily under a low applied stress. Therefore, it is likely that α'' can be stress-induced at a very low stress level in the rapidly swaged rods and the tangent modulus decreases in an initial linear range of the stress-strain curve at cycle 1.

Adiabatic temperature increase and its effect on stress-induced α'' transformation were further investigated using a deformation simulator. Temperature changes during compression were measured with a thermo-couple embedded in a cylindrical sample and recorded in a data logger. Fig. 18 shows the maximum temperature reached during compression. Deformation at a strain rate of 10 s^{-1} was impossible in the larger sample B because of the operation of an over-loaded system under a high strain rate. The maximum temperature by adiabatic heating was sensitive to the sample size, applied strain and strain rate. In sample A the maximum temperature increased with increasing strain rate and reached approximately 450 K at strain rates of 1 s^{-1} and 10 s^{-1} . The OM micrographs of sample B deformed at 1 s^{-1} and 10^{-3} s^{-1} are shown in Fig. 19. Deformation-induced α'' plates can be seen for both strain rates. Fig. 20 shows the XRD profiles on the section perpendicular to the compression direction of sample B deformed at 1 s^{-1} and 10^{-3} s^{-1} . Although both profiles are similar to each other, it is obvious

that the relative intensity of $(220)_{\alpha''}$ to $(211)_{\beta}$ is stronger at 10^{-3}s^{-1} than at 1s^{-1} . This result suggests that stress-induced α'' at 1s^{-1} reverts to β to some extent due to adiabatic heating. Thus, the results obtained from the experiment using a deformation simulator are consistent with the conclusions drawn from the swaging studies described above.

In this study the Young's modulus of the hot-forged/quenched rod with a strong $\langle 110 \rangle_{\beta}$ texture along the forging direction was found to be 56 GPa, which was close to the Young's modulus of 56.3 GPa in the $\langle 110 \rangle$ oriented $\beta\text{Ti-24Nb-4Zr-8Sn}$ single crystal (Zhang et al., 2011). Therefore, it is noted that hot-forging to develop a $\langle 110 \rangle_{\beta}$ texture is a practical and economical process in β Ti alloys to achieve a low Young's modulus of approximately 56 GPa. The Young's modulus further decreased to 44 GPa by rapid swaging and 45 GPa by gradual swaging through α'' formation, because a large amount of α'' martensite with a $\langle 010 \rangle_{\alpha''}$ texture was retained after swaging (Figs. 5 and 6). Assuming that the rule of mixture is applicable to Young's modulus of Ti-33.6Nb-4Sn with strong $(\beta+\alpha'')$ textures in which α'' is completely stabilized by severe rolling and gradual swaging, the Young's modulus in $\langle 010 \rangle_{\alpha''}$ has been estimated to be 38 GPa (Hanada et al., 2013). Therefore, both the α'' formation and the $\langle 010 \rangle_{\alpha''}$ texture development contribute to the decrease in the Young's modulus by swaging. Thus, it is concluded that the Young's modulus of β Ti-33.6Nb-4Sn can be reduced by texture control of β and α'' phases.

4. Summary

The effect of swaging at two diameter reduction rates on Young's modulus in β Ti-33.6Nb-4Sn rods was investigated by X-ray diffraction, thermography, microstructural observations, deformation simulator analysis and cyclic tensile deformation. The obtained results are summarized as follows.

1. Stress-induced α'' martensite is stabilized by swaging and its volume fraction was greater in a gradually swaged rod than in a rapidly swaged rod.
2. Thermography revealed that the rapidly swaged rod was adiabatically heated over 500 K.
3. Deformation simulator analysis indicated that a compressed sample was adiabatically heated and the maximum temperature increased with increasing strain rate.
4. Young's modulus, which was determined from a linear slope of a stress-strain curve, decreased from 56 GPa in the hot forged/quenched rod to 44 GPa in the rapidly 75% swaged rod and to 45 GPa in the 75% gradually swaged rod.
5. Tangent modulus, which was determined by the slope of the tangent to any point on the stress-strain curve, decreased with strain even in the initial linear range in the hot forged/quenched rod and the rapidly swaged rod, while it was mostly unchanged in the

gradually swaged rod.

6. Tangent moduli in swaged β Ti-33.6Nb-4Sn rods were affected by the stabilization of α'' .

7. Low Young's modulus of 45 GPa and high strength over 800MPa were obtained by gradual swaging to the 75% reduction in diameter.

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Captions

- Fig. 1 Optical micrographs of hot-forged/quenched Ti-33.6Nb-4Sn in (a) a transverse section and (b) a longitudinal section.
- Fig. 2 DSC curves of hot-forged/quenched Ti-33.6Nb-4Sn in (a) cooling and (b) heating.
- Fig. 3 Adiabatic temperature increase in Ti-33.6Nb-4Sn swaged rapidly and gradually.
- Fig. 4 Optical micrographs after (a, b) gradual swaging and (c, d) rapid swaging in the (a, c) transverse sections and (b, d) longitudinal sections.
- Fig. 5 XRD patterns in the transverse section of (a) hot-forged/quenched, (b) rapidly swaged and (c) gradually swaged Ti-33.6Nb-4Sn.
- Fig. 6 $(110)_\beta$ pole figures of (a) gradually swaged and (b) rapidly swaged Ti-33.6Nb-4Sn and $(020)_{\alpha'}$ pole figures of (c) gradually swaged and (d) rapidly swaged Ti-33.6Nb-4Sn.
- Fig. 7 Tensile stress-strain curves of hot-forged/quenched, rapidly swaged and gradually swaged Ti-33.6Nb-4Sn.
- Fig. 8 Cyclic stress-strain curves of hot-forged/quenched Ti-33.6Nb-4Sn. (a) cycle 1, (b) cycle 2, (c) cycle 5 and (d) cycle 10.
- Fig. 9 Stress-strain curves on loading at cycles 1 and 2 of hot-forged/quenched Ti-33.6Nb-4Sn.
- Fig. 10 Tangent modulus at cycles 1 and 2 of hot-forged/quenched Ti-33.6Nb-4Sn.
- Fig. 11 Cyclic stress-strain curves of rapidly swaged Ti-33.6Nb-4Sn. (a) cycle 1, (b) cycle 2, (c) cycle 5 and (d) cycle 10.
- Fig. 12 Stress-strain curves on loading at cycles 1 and 2 of rapidly swaged Ti-33.6Nb-4Sn.
- Fig. 13 Tangent modulus at cycles 1 and 2 of rapidly swaged Ti-33.6Nb-4Sn.
- Fig. 14 Cyclic stress-strain curves of gradually swaged Ti-33.6Nb-4Sn. (a) cycle 1, (b) cycle 2, (c) cycle 5 and (d) cycle 10.
- Fig. 15 Tangent modulus at cycles 1 and 2 of gradually swaged Ti-33.6Nb-4Sn.
- Fig. 16 Residual strain of hot-forged/quenched, rapidly and gradually swaged

Ti-33.6Nb-4Sn as a function of deformation cycle.

Fig. 17 A DSC curve in heating of gradually swaged Ti-33.6Nb-4Sn.

Fig. 18 Maximum temperature reached during deformation by a deformation simulator as a function of strain rate.

Fig. 19 Optical micrographs of Ti-33.6Nb-4Sn solution treated and deformed by a deformation simulator at strain rates of (a) 1 s^{-1} and (b) 10^{-3} s^{-1} .

Fig. 20 XRD patterns in the transverse section of Ti-33.6Nb-4Sn solution treated and deformed by a deformation simulator at strain rates of (a) 1 s^{-1} and (b) 10^{-3} s^{-1} .

Fig. 1

Fig. 2

Fig. 3

Fig. 4

Fig. 5

Fig. 6

	Intensity	Color
1	12583.604	
3	25769.291	
5	38954.978	
7	52140.665	
10	71919.196

	Intensity	Color
1	13945.850	
3	29700.730	
5	45455.610	
7	61210.490	
10	84842.810

	Intensity	Color
1	9447.504	
3	26016.891	
5	42586.278	
7	59155.665	
10	84009.746

	Intensity	Color
1	4851.533	
3	12090.578	
5	19329.624	
7	26568.669	
10	37427.237	

Fig. 7

Fig. 8

Fig. 9

Fig. 10

Fig. 11

Fig. 12

Fig. 13

Fig. 14

Fig. 15

Fig. 16

Fig. 17

Fig. 18

Fig. 19

Fig. 20