Effect of Oxygen Addition on the Formation of α'' Martensite and Athermal ω in Ti-Nb Alloys

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The effect of the addition of oxygen on the formation of the microstructure quenched from the β phase of Ti-Nb alloy was investigated based on the solid solution treatment (SST) temperature in the β phase. The alloy ingots of Ti-(12, 14, and 18)-at% Nb-(0, 1, and 3)-at% O were arc-melted. They were homogenized at 1200°C for 3.6 ks and then hot-rolled at 850°C into 1.5-mm thick sheets. The specimens were solution-treated at 1050 to 1200°C for 0.6 ks and then quenched in iced brine. The microstructure of the Ti-(12, 14, and 18)-at% Nb alloys exhibited an α'' phase regardless of the SST temperature. The addition of oxygen in the Ti-(14 and 18)-at% Nb alloys suppressed the $\beta \rightarrow \alpha''$ martensitic transformation; therefore, the quenched structure became the $\beta + \omega_a$ phase. The further addition of oxygen suppressed the $\beta \rightarrow \omega_a$ transformation during quenching. The effect of oxygen addition on the phase transformations of $\beta \rightarrow \alpha''$ and $\beta \rightarrow \omega_a$ during quenching from the β phase was weakened with an increasing SST temperature. [doi:10.2320/matertrans.ME201908]

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1. Introduction*

Because titanium alloys exhibit excellent properties, such as high specific strength, corrosion resistance, and heat resistance, they are widely used in aircrafts, aerospace industries, automobile, and medical applications.^{1–5)} Microstructure control utilizing phase transformations is often required to adjust the mechanical properties, such as strength and ductility, when the titanium alloys are applied as a structural component. The microstructure formed by quenching from the β phase in a titanium alloy varies by increasing the amount of isomorphous-type β phase stabilizers, such as niobium and molybdenum as follows: α' martensite, α'' martensite, $\beta + \omega_a$ (athermal ω), and β (no phase transformation). The microstructure differences influence the mechanical properties, such as the Young's modulus and hardness.^{6,7)} If an α -phase stabilizing element, such as aluminum or tin, is added to Ti-Nb or Ti-Mo alloys that exhibit $\beta \rightarrow \alpha''$ martensitic transformation, the $\beta \rightarrow \alpha''$ martensitic transformation is suppressed.^{8–11)} Ikeda *et al.* investigated the effect of an aluminum addition on the formation of the microstructure quenched from the β phase in the Ti-11-mass% Mo alloy and clarified that the $\beta \rightarrow \alpha''$ martensitic transformation was suppressed by a 3-mass% Al addition.¹²⁾ The suppression of the $\beta \rightarrow \alpha''$ martensitic transformation by the aluminum addition was caused by the lattice distortion in the β phase, and the aluminum addition would increase resistance for lattice displacement of the martensitic transformation. The $\beta \rightarrow \alpha''$ martensitic transformation could also be suppressed by the addition of oxygen, which is another α -phase stabilizing element.^{13–17)} Tahara *et al.* clarified that the $\beta \rightarrow \alpha''$ martensitic transformation of the Ti-20-at% Nb alloy was suppressed by a 0.7-at% oxygen addition. They suggested that the formation of the nano-domain structure induced by the strain field of oxygen atoms caused the suppression of the martensitic

transformation owing to the oxygen addition.¹⁸⁾ Although the effect of the oxygen addition on the formation of the microstructure quenched from the β phase was clarified, the details, such as the dependency of the solid solution treatment (SST) temperature on the effect of the oxygen addition, are unknown.

In this study, the effect of the oxygen addition on the formation of the microstructure quenched from the β phase in Ti-Nb alloy was evaluated based on the difference in the SST temperature in the β phase.

2. Experimental Procedures

Sponge titanium (>99 mass%), granular niobium (>99.9 mass%), TiO powder (99.9 mass%), and titanium foil (>99.5 mass%) were used for the sample preparation. These pure metals and oxides were weighed for a nominal alloy composition of Ti-(12, 14, and 18)at% Nb-(0, 1, and 3) at% O. The TiO powder was wrapped in titanium foil during weighing. A button-shaped ingot of 10 g was arc-melted using a tungsten electrode in a water-cooled copper hearth. A pure titanium block was melted to purify the argon gas atmosphere before argon arc melting. The alloy button was melted eight times and was flipped over before each melting run to create a uniform composition. The niobium and oxygen concentrations of the alloys were measured by energy dispersive spectroscopy (JEOL: JEM-6310) and the infrared absorption method (LECO: ONH836), respectively. The ingot was wrapped with tantalum foil, and after evacuating, it was sealed in a quartz tube filled with argon at 0.02 MPa. It was homogenized at 1200°C for 3.6 ks in the β phase and then quenched into iced brine. The homogenized alloy button was hot-rolled at 850°C to obtain an approximately 1.5-mm thick plate sample. The plate sample was cut into small platelets of $10 \times 10 \times 1.5 \text{ mm}^3$, and the samples were subjected to a SST at temperatures of 1050 to 1200°C for 0.6 ks under argon gas flow and then were quenched in iced brine. In order to analyze the microstructure of the heat-treated sample, the sample was etched with an 8% HF and

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20% HNO₃ aqueous solution and observed using an optical microscope and scanning electron microscope. X-ray diffractometry (XRD) was used to identify the phase of the sample. For a quantitative evaluation of the grain size of the β phase and the amount of α'' phase in the β phase, image processing software, *Image J*,¹⁹⁾ was used.

3. Experimental Results

3.1 Influence of the oxygen addition on the phase transformation during quenching from 1050°C in the β phase

The concentration of niobium in the alloys used in this study was within ± 0.2 at% from the nominal composition. Table 1 lists the measurement results of the oxygen concentration in the alloys. The oxygen-free alloys, Ti–(12, 14, and 18)-at% Nb contained approximately 0.25 at% (0.075 mass%) oxygen; however, in this study, they are displayed as Ti–(12, 14, and 18)-at% Nb–0 at% O or Ti–(12, 14, and 18)at% Nb because oxygen was not intentionally added. Because the oxygen content of each alloy is close to the nominal composition, the oxygen concentration of an alloy is displayed in the nominal composition hereafter. The nitrogen and hydrogen concentrations in the alloys were approximately 130 ppm and 40 ppm in weight percent,

respectively. Hereafter, the composition will be displayed in the atomic percent, and the unit will be omitted.

Figures 1 and 2 show the optical micrographs of Ti–12Nb–(0, 1, and 3)O alloys quenched from 1050°C in the β phase and the X-ray diffraction profiles obtained from these alloys, respectively. As shown in Fig. 1(a), fine plates were formed in the equiaxed grains of approximately 300 μ m in the Ti–12Nb alloy. The XRD profile in Fig. 2(a) shows that the microstructure of the Ti–12Nb alloy consisted of the α'' phase. Therefore, the fine plates were α'' phase, and the equiaxed grains corresponded to the prior β phase. As shown in Figs. 1(b) and (c) and Figs. 2(b) and (c), the Ti–12Nb–(1 and 3)O alloys also exhibited fine plates of the α'' phase, which were observed in the Ti–12Nb alloy. The detailed analysis of the diffraction peak position of the orthorhombic α'' phase in Fig. 2 revealed that the diffraction peak position changed with the addition of oxygen in the Ti–12Nb alloy. The 020 and 002 diffraction peak positions relating to the b- and c-axes of the orthorhombic α'' phase shifted to a higher diffraction angle by the addition of oxygen. However, the 110, 111, and 112 diffraction peak positions, including the plane index of the a-axis, shifted to a lower diffraction angle. These results indicated that the a-axis of the α'' phase was expanded by the oxygen addition, while the contraction of the b- and c-axes of the α'' phase occurred. The grain size of the prior β phase decreased with an increase in the oxygen concentration in the alloys. The average grain sizes of the prior β phase in the Ti–12Nb–(0, 1, and 3)O alloy were 310, 160, and 100 μ m, respectively. This implied that some of the

Table 1 Oxygen concentration of each alloy.

Nominal alloy composition [at%]	Oxygen concentration measured [at%]
Ti–12Nb	0.24
Ti–12Nb–1O	1.22
Ti–12Nb–3O	3.17
Ti–14Nb	0.29
Ti–14Nb–1O	1.21
Ti–14Nb–3O	3.11
Ti–18Nb	0.25
Ti–18Nb–1O	1.21
Ti–18Nb–3O	3.22

Fig. 2 X-ray diffraction profiles of the (a) Ti–12Nb, (b) Ti–12Nb–1O, and (c) Ti–12Nb–3O alloys quenched from 1050°C. The solid circles represent the diffraction peaks of the α'' phase.

Fig. 1 Optical micrographs for the (a) Ti–12Nb, (b) Ti–12Nb–1O, and (c) Ti–12Nb–3O alloys quenched from 1050°C.

Fig. 3 Optical micrographs for the (a) Ti-14Nb, (b) Ti-14Nb-1O, and (c) Ti-14Nb-3O alloys quenched from 1050°C. The inserts in (a)–(c) are magnified images obtained by scanning electron microscopy. The acronym G.B. in the insert in (c) is an abbreviation of grain boundary.

Fig. 4 X-ray diffraction profiles of the (a) Ti-14Nb, (b) Ti-14Nb-1O, and (c) Ti-14Nb-3O alloys quenched from 1050°C. The solid circles, open squares, and solid triangles represent the diffraction peaks of the α'' , β , and ω_a phases, respectively.

oxygen atoms were segregated at the grain boundaries of the β phase preventing grain boundary migration by the solute drag effect. Figures 3 and 4 show the optical micrographs of the Ti-14Nb-(0, 1, and 3)O alloys quenched from 1050°C in the β phase and the X-ray diffraction profiles obtained from these alloys, respectively. The inserts in Fig. 3 are a magnified image of each micrograph observed by the scanning electron microscope. Figures 3(a) and 4(a) show that the Ti-14Nb alloy exhibited fine plates of the α'' phase in the equiaxed prior β grains. From Figs. 3(b) and 4(b), the addition of 1 at% oxygen to the Ti-14Nb alloy did not change the microstructure, except for a decrease of the prior β grain size. The α'' martensite plates were partially formed in the β grains in the Ti-14Nb-3O alloy, as shown in Fig. 3(c). The

α'' martensite plates were mainly nucleated from the grain boundaries of the β phase, as shown in the inset of Fig. 3(c). The martensite plates of the α'' phase were refined by the 3 at% oxygen addition. Figure 4(c) shows that the constituent phases of the Ti-14Nb-3O alloy were the α'' and ($\beta + \omega_a$) phases, suggesting that the area without the martensite plates in Fig. 3(c) contained ($\beta + \omega_a$) phases. The average grain sizes of the β phase in the Ti-14Nb-(0, 1, and 3)O alloys were 290, 140, and 110 μm , respectively, suggesting that grain growth of the β phase was suppressed by the oxygen addition. The optical micrographs of the Ti-18Nb-(0, 1, and 3)O alloys quenched from 1050°C in the β phase and the X-ray diffraction profiles obtained from these alloys are shown in Figs. 5 and 6, respectively. Figures 5(a) and 6(a) show that the microstructure of the Ti-18Nb alloy exhibited fine plates of the α'' phase. The Ti-18Nb-(1 and 3)O alloys exhibited β grains without α'' martensite plates, and they showed diffraction peaks of the β and ω_a phases in the XRD profiles in Fig. 6(b) and (c). The intensities of the diffraction peaks of the ω_a phase in the Ti-18Nb-3O alloy were weaker than those in the Ti-18Nb-1O alloy, implying that the addition of oxygen suppressed the formation of the ω_a phase. The average grain sizes of the β phase in the Ti-18Nb-(0, 1, and 3)O alloys were 280, 130, and 100 μm , respectively. Thus, the oxygen atoms were segregated at the grain boundaries and inhibited grain growth during the SST. Figure 7 shows the volume fraction of the α'' phase formed by quenching from the β phase in the Ti-Nb alloys containing 0, 1, and 3 at% O as a function of the niobium content. This figure also shows the results for the Ti-(11, 13, 15, 16, 17, and 20)Nb-(0, 1, and 3)O alloys, which are not described in this study. The volume fraction of the α'' phase was 100% in all Ti-Nb binary alloys in the composition range of 11 to 20 at% Nb. Conversely, in the alloys containing 1 or 3 at% O, the volume fraction of the α'' phase

Fig. 5 Optical micrographs for (a) Ti-18Nb, (b) Ti-18Nb-1O, and (c) Ti-18Nb-3O quenched from 1050°C.

addition on the $\beta \rightarrow \alpha''$ transformation was not observed in the Ti-12Nb-(0, 1, and 3)O alloys. Here, $\Delta E_f^{\alpha''-\beta}$ does not sufficiently represent the effect of the oxygen addition on the $\beta \rightarrow \alpha''$ transformation. Uesugi *et al.*²¹⁾ recently calculated the Gibbs free energy of the β and α'' phases in Ti-(0~38.9)Nb-(0 and 2.7)O at various temperatures from 0 to 1300°C. They showed that at a low Nb concentration less than approximately 20 at% Nb with a high Ms temperature, the oxygen addition increased the transformation temperature of $\beta \rightarrow \alpha''$, i.e., the driving force for the $\beta \rightarrow \alpha''$ transformation increased and promoted the $\beta \rightarrow \alpha''$ transformation. However, they also showed that at a high Nb concentration greater than approximately 20 at% Nb with a low Ms temperature, the oxygen addition lowered the transformation temperature of $\beta \rightarrow \alpha''$. Based on their results, the oxygen addition could promote the $\beta \rightarrow \alpha''$ transformation owing to the chemical driving force; however, our experimental results showed that the oxygen addition in the Ti-(14 and 18)Nb alloy suppressed the $\beta \rightarrow \alpha''$ transformation. This implies that the influence on the chemical free energy change owing to the oxygen addition is not the main cause of the suppression of the $\beta \rightarrow \alpha''$ martensitic transformation. Because the oxygen atoms are an interstitial element and induce local distortion of the β -phase lattice, the oxygen addition would inhibit the collective displacement of the atoms required for martensitic transformation, resulting in an increase of the strain energy of the martensitic transformation. Therefore, the formation of the strain field owing to the oxygen addition should be the main factor for the suppression of the $\beta \rightarrow \alpha''$ martensitic transformation.^{14-18,20,22)} As shown in Fig. 3, the plates of α'' martensite were refined by the oxygen addition, suggesting that the growth of the martensite plates was inhibited by the strain field of the oxygen atoms in the β grains.

The α'' martensite phase is often nucleated from the lattice defects, such as the grain boundary, including large and small angle grain boundaries. The nucleation of martensite on the grain boundary could be suppressed by the addition of oxygen. Because the oxygen atom has a smaller radius of 60 pm than that of the titanium atom of 140 pm, it occupies the interstitial lattice sites and grain boundaries. As shown in Figs. 1, 3, and 5, the β grain size after the SST at 1050°C decreased owing to the addition of oxygen, indicating that segregation of oxygen at the grain boundary occurred and inhibited grain boundary migration. The equilibrium segregation concentration of oxygen on the grain boundary is given by the Seah and Hondros' segregation equation,²³⁻²⁵⁾ as shown in eq. (1).

$$x_b = x_{b0} \left(\frac{x_m}{x_{m0} - x_m} \right) / \left(\frac{1}{K} + \frac{K-1}{K} \frac{x_m}{x_{m0}} \right) \quad (1)$$

Here, x_{b0} and x_{m0} are the saturated oxygen concentrations on the grain boundaries at a certain temperature T and the maximum concentration of solute oxygen in the intragranular parent phase, i.e., the solid solubility limit, respectively. The x_m is the oxygen concentration in the intragranular parent phase, which is determined from the alloy composition. The temperature factor, K , is given by $K = \exp(E/RT)$, where E and R are the grain boundary segregation energy and the gas constant, respectively. Syarif *et al.*²⁶⁾ evaluated the segrega-

tion amount of oxygen on the grain boundary at 1000°C for the Ti-10Mo alloy based on eq. (1). They calculated the concentration of oxygen segregation on the grain boundaries as $x_b = 0.45$ by substituting the following values: $T = 1273 \text{ K}$, $x_{b0} = 0.67$, $x_{m0} = 0.072$, $E = 7320 \text{ J/mol}$, and $x_m = 0.022$ (this value is estimated from Table 2 in their study). This calculation indicated that up to 45 at% of oxygen is concentrated on the grain boundary when the oxygen concentration in the parent phase is 2.2 at%. This segregation of oxygen to the grain boundary has also been reported by Wu *et al.*²⁷⁾ and Wei *et al.*²⁸⁾ Although it is difficult to apply their data to evaluate the grain boundary segregation in the Ti-Nb alloy system, the tendency for oxygen segregation on the grain boundary was qualitatively evaluated. Because the alloying elements of molybdenum and niobium in a Ti alloy are isomorphous type β phase stabilizers, a qualitative estimation of the grain boundary segregation of oxygen in the Ti-Nb system could be performed using their data. If the amount of oxygen in the parent β phase is 1 at% or 3 at%, the oxygen concentration on the grain boundary should be 19 at% or 68 at%, respectively, based on eq. (1). Thus, an increase of 2 at% oxygen in the parent β phase results in an increase in the amount of oxygen on the grain boundary by approximately 50 at%. The segregation of oxygen on the grain boundary should lower the grain boundary energy, suppressing the nucleation of α'' martensite on the grain boundary. The microstructure difference between the Ti-14Nb-1O and Ti-14Nb-3O alloys could be from the large difference of oxygen concentration on the grain boundary. In contrast to the Ti-14Nb alloy, the $\beta \rightarrow \alpha''$ martensitic transformation of the Ti-18Nb alloy was fully suppressed by the addition of 1 at% oxygen. This is because the chemical driving force of the $\beta \rightarrow \alpha''$ martensitic transformation in the Ti-18Nb alloy would be lower than that in the Ti-14Nb alloy. However, the chemical driving force of the $\beta \rightarrow \alpha''$ martensitic transformation of the Ti-12Nb alloy would be greater than those of the Ti-14 and 18Nb alloys. Although the oxygen addition in the Ti-12Nb alloy increased the elastic strain energy for the formation of α'' martensite, the absolute value of the chemical driving force was greater than that of the elastic strain energy; therefore, the suppression of α'' martensite by the oxygen addition was not observed in the Ti-12Nb-(1 and 3)O alloys.

As shown in Fig. 6, the oxygen addition in the Ti-18Nb alloy inhibited the $\beta \rightarrow \omega_a$ transformation as well. Because the $\beta \rightarrow \omega_a$ transformation formed fine ω_a area or particles within the β grain uniformly, the effect of the oxygen addition on the suppression of the $\beta \rightarrow \omega_a$ transformation was not owing to the oxygen segregation at the grain boundary. The crystal structure change from the β to ω_a phase caused the {111} planes of the β phase to move closer together in the $\langle 111 \rangle$ direction.²⁹⁾ This displacement of the atomic planes should be inhibited by the strain field of the lattice by the oxygen addition; therefore, suppression of the $\beta \rightarrow \omega_a$ transformation occurred owing to the oxygen addition.

4.2 Influence of the SST temperature on the effect of the oxygen addition on the $\beta \rightarrow \alpha''$ martensitic transformation

As shown in Fig. 12, the effect of the oxygen addition on

the suppression of the $\beta \rightarrow \alpha''$ martensitic transformation was weakened by an increasing SST temperature. The increase of the SST temperature increased the number of vacancies in the β grain,³⁰⁾ which should weaken the effect of the oxygen addition on the $\beta \rightarrow \alpha''$ martensitic transformation. Oxygen atoms occupy the interstitial atomic sites and cause lattice expansion,¹⁸⁾ while the vacancies should induce lattice contraction. Therefore, the strain fields generated by the oxygen and vacancies would relax when they are located close to each other. Because the number of vacancies increases with an increasing SST temperature, relaxation of the strain field of the oxygen atom by a combination with the vacancy should be enhanced at a higher SST temperature, weakening the effect of the oxygen addition on the suppression of the $\beta \rightarrow \alpha''$ martensitic transformation. To verify this qualitative analysis, a quantitative evaluation should be performed. As shown in Figs. 3(b) and (c), the α'' martensite phase was partially formed in the Ti-14Nb-3O alloy after the SST at 1050°C, while reducing the oxygen content of 2 at%, i.e., in the Ti-14Nb-1O alloy, the α'' martensite phase was fully formed. Conversely, by changing the SST temperature of the Ti-14Nb-3O alloy from 1050°C to 1200°C, the α'' martensite phase was fully formed, as shown in Figs. 10(a) and (c). As an hypothesis, the strain field caused by the increase of 2 at% oxygen could be relaxed by the vacancies introduced from changing the SST temperature from 1050°C to 1200°C. From the theoretical equation for the equilibrium vacancy concentration,³¹⁾ the number of vacancies at temperatures of 1050°C and 1200°C are calculated as 617 and 2647 in 10^9 atoms of the β phase using the formation energy of the vacancy for pure titanium, 1.63 eV.³²⁾ The increase in the number of vacancies is 2030, while the number of oxygen atoms for 2.0 at% O in 10^9 atoms is 2×10^7 . For the above hypothesis, one vacancy should relax the strain field of approximately 9850 oxygen atoms, implying the hypothesis mentioned above is not accurate.

The oxygen concentration of the grain boundary should change with an increasing SST temperature. When the oxygen concentration is not high, eq. (1) can be approximated by eq. (2).

$$x_b = x_{b0} \left(\frac{x_m}{x_{m0}} \right) K \quad (2)$$

As shown in eq. (2), the oxygen concentration at the grain boundary, x_b , will decrease with an increasing temperature if the value of E is positive in the temperature factor, $K = \exp(E/RT)$. In addition, because the solubility limit in the matrix, x_{m0} , increases as the temperature rises, which can be confirmed in the Ti-O phase diagram, the oxygen concentration at the grain boundary, x_b , should decrease with an increasing temperature. The decrease in oxygen segregation on the grain boundary owing to the increase in the SST temperature weakens the effect of the oxygen addition on the suppression of the α'' martensite transformation.

As mentioned in the introduction, Tahara *et al.*¹⁸⁾ proposed that the formation of nano-domains induced by the strain field of the oxygen atoms suppressed the $\beta \rightarrow \alpha''$ martensitic transformation. In addition, Ishiguro *et al.*³³⁾ suggested that the formation of the nano-domains are related to the spinodal decomposition of the β phase in the Ti-Nb-O system. The

experimental results in Fig. 10 can be explained as follows. If the Ti-14Nb-3O alloy at 1050°C is in the region of spinodal decomposition, nano-domains would form during quenching to suppress the $\beta \rightarrow \alpha''$ martensitic transformation. Conversely, if the Ti-14Nb-3O alloy at 1200°C is out of the region of spinodal decomposition, no nano-domains form during quenching, i.e., $\beta \rightarrow \alpha''$ martensitic transformation should occur during quenching. Therefore, another explanation for the influence of the SST temperature on the effect of the oxygen addition on the $\beta \rightarrow \alpha''$ martensitic transformation could be owing to the nano-domain formation and spinodal decomposition of the β phase. A further study on the microstructure in the nano-scale using transmission electron microscopy is required and is under investigation.

5. Conclusions

The influence of the oxygen addition on the formation of the microstructure in Ti-Nb alloys quenched from the β phase was examined, and the following conclusions were obtained.

- (1) Although the microstructure of the Ti-14Nb alloy quenched from 1050°C in the β phase was the α'' phase, the $\beta \rightarrow \alpha''$ martensitic transformation was suppressed by the addition of 3-at% O. This was confirmed in the Ti-18Nb alloy, and the α'' martensitic transformation was fully suppressed by the addition of 1-at% O. The suppression of the $\beta \rightarrow \alpha''$ martensitic transformation owing to the oxygen addition was caused by the following factors: (i) the oxygen atoms induced lattice distortion which inhibited the shear deformation required for the $\beta \rightarrow \alpha''$ martensitic transformation, and (ii) the oxygen atoms segregated at the grain boundary reducing the grain boundary energy. Therefore, nucleation of the α'' martensite phase at the grain boundary was suppressed.
- (2) The effect of the oxygen addition on the $\beta \rightarrow \alpha''$ martensitic transformation in the Ti-Nb alloys was weakened by an increase of the SST temperature. The segregation of oxygen on the grain boundary was decreased with an increasing SST temperature, weakening the effect of the oxygen addition on the $\beta \rightarrow \alpha''$ martensitic transformation.
- (3) The formation of ω_a in the β phase instead of α'' martensite was confirmed in the Ti-18Nb-1O alloy. The amount of ω_a decreased in the Ti-18Nb-3O alloy. In addition, the oxygen addition suppressed the $\beta \rightarrow \omega_a$ transformation. The inhibition of the atomic plane displacement required for the $\beta \rightarrow \omega_a$ transformation was induced by the strain field of the oxygen atoms.

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REFERENCES

- 1) I. Inagaki, T. Takechi, Y. Shirai and N. Ariyasu: Nippon Steel Sumitomo Met. Tech. Rep. **106** (2014) 22–27.
- 2) F.H. Froes, H. Friedrich and D. Bergoint: *JOM* **56** (2004) 40–44.
- 3) D. Banerjee and J.C. Williams: *Acta Mater.* **61** (2013) 844–879.
- 4) L. Le Guéhennec, A. Soueidan, P. Layrolle and Y. Amouriq: *Dent. Mater.* **23** (2007) 844–854.
- 5) S. Nishikiori: *J. JILM* **55** (2005) 557–560.
- 6) S. Kobayashi, K. Manabe, A. Nakao and S. Okano: *Titan. Japan* **66** (2018) 31–37.
- 7) C.M. Lee, C.P. Ju and J.H.C. Lin: *J. Oral Rehabil.* **29** (2002) 314–322.
- 8) M. Ijiri, Y. Tomita, T. Ishikawa, K. Kadowaki and Y. Takemoto: *J. Japan Inst. Met. Mater.* **81** (2017) 345–351.
- 9) B.L. Wang, Y.F. Zheng and L.C. Zhao: *Mater. Sci. Eng. A* **486** (2008) 146–151.
- 10) T. Maeshima and M. Nishida: *Mater. Trans.* **45** (2004) 1096–1100.
- 11) L.A. Matlakhova, A.N. Matlakhov, S.N. Monteiro, S.G. Fedotov and B.A. Goncharenko: *Mater. Sci. Eng. A* **393** (2005) 320–326.
- 12) M. Ikeda, S. Komatsu and T. Sugimoto: *J. JILM* **50** (2000) 435–440.
- 13) H. Kim, H. Hosoda and S. Miyazaki: *J. JILM* **55** (2005) 613–617.
- 14) J.I. Kim, H. Kim, H. Hosoda and S. Miyazaki: *Mater. Trans.* **46** (2005) 852–857.
- 15) M. Tahara, T. Inamura, H. Kim, H. Hosoda and S. Miyazaki: *J. JILM* **62** (2012) 257–262.
- 16) M. Tahara, H. Kim, T. Inamura, H. Hosoda and S. Miyazaki: *Acta Mater.* **59** (2011) 6208–6218.
- 17) H. Kim: *Materia Japan* **53** (2014) 11–17.
- 18) M. Tahara, T. Inamura, H. Kim, S. Miyazaki and H. Hosoda: *Scr. Mater.* **112** (2016) 15–18.
- 19) C.A. Schneider, W.S. Rasband and K.W. Eliceiri: *Nat. Methods* **9** (2012) 671–675.
- 20) R. Salloom, D. Reith, R. Banerjee and S.G. Srinivasan: *J. Mater. Sci.* **53** (2018) 11473–11487.
- 21) T. Uesugi, J. Shimamoto, D. Minami, Y. Takigawa and K. Higashi: *Taikai Kouen Gaiyou*, (Keikinzoku Gakkai, 2018) pp. 341–342.
- 22) Y. Nii, T. Arima, H.Y. Kim and S. Miyazaki: *Phys. Rev. B* **82** (2010) 214104.
- 23) M.P. Seah and E.D. Hondros: *Proc. R. Soc. London* **335** (1973) 191–212.
- 24) M.P. Seah and E.D. Hondros: *Scr. Metall.* **7** (1973) 735–737.
- 25) E.D. Hondros and M.P. Seah: *Metall. Mater. Trans. A* **8** (1977) 1363–1371.
- 26) J. Syarif, T.N. Rohmannudin, M.Z. Omar, Z. Sajuri and S. Harjanto: *J. Min. Metall. Sect. B Metall.* **49** (2013) 285–292.
- 27) X. Wu, J. del Prado, Q. Li, A. Huang, D. Hu and M.H. Loretto: *Acta Mater.* **54** (2006) 5433–5448.
- 28) Q. Wei, L. Wang, Y. Fu, J. Qin, W. Lu and D. Zhang: *Mater. Des.* **32** (2011) 2934–2939.
- 29) Z. Nishiyama: *Bulletin of the Japan Institute of Metals* **14** (1975) 499–512.
- 30) Y. Miura: *Materia Japan* **45** (2006) 283–288.
- 31) D.R. Askeland and P.P. Phulé: *Essentials of Materials Science and Engineering 1 Edition*, (CL Engineering, London, 2004).
- 32) A.V. Ruban, H.L. Skriver and B. Johansson: *Phys. Rev. B* **59** (1999) 11693–11703.
- 33) Y. Ishiguro, Y. Tsukada and T. Koyama: *Comput. Mater. Sci.* **151** (2018) 222–230.