Effects of Heat Treatment on Mechanical Properties and Corrosion Fatigue Strength in Physiological Saline Solution of New Titanium Alloys for Medical Implants

Yoshimitsu Okazaki, Yoshimasa Ito* and Tetsuya Tateishi**

Mechanical Engineering Laboratory, Agency of Industrial Science and Technology, Ministry of International Trade and Industry, 1-2 Namiki, Tsukuba 305, Japan

* Technical Development Group Kobe Steel, Ltd, Wakinohamacho, Kobe 651, Japan

** National Institute for Advanced Interdisciplinary Research, 1-1 Higashi, Tsukuba 305, Japan

Ti-15mass%Zr-4%Nb-4%Ta-0.2%Pd and Ti-15%Sn-4%Nb-2%Ta-0.2%Pd alloys not containing V and Al for surgical implantation were cast by plasma electron beam melting. The effect of heat treatment on room temperature strength and corrosion fatigue strength were investigated by aging treatment after solution treatment. The alloys were forged in their α and $\alpha + \beta$ regions, and then solution-treated at 953–1173 K for 3.6 ks. Optimized solution treatment conditions are 1038 K -3.6 ks for Ti-Zr alloy and 1098 K -3.6 ks for Ti-Sn alloy containing 15–20 vol% of primary α phase. After solution treatment under the optimized conditions, aging treatment was carried out at 623–873 K for 600 s ~ 108 ks. Optimized aging conditions for Ti-Zr and Ti-Sn alloys were at 673 K for 36 ks and at 723 K for 36 ks, respectively. The room temperature strengths of these solution treated and aged alloys were about 30% higher than the annealed alloys. It is revealed by transmission electron microscopic observation that, fine α phase precipitates in α' martensitic matrix are responsible for the age hardening. The corrosion fatigue test in a physiological saline solution was carried out under the condition of a tension to tension mode with a sine wave at a stress ratio of 0.1, at a frequency of 10 Hz in Eagle's MEM+fetal bovine serum solution using an environment cell with 90%N₂+5% CO₂+5% O₂ gas bubbling at 310 K. The fatigue strength at 10⁷ cycles in solution-treated and aged Ti-15%Zr-4%Nb-4%Ta-0.2%Pd and Ti-15%Sn-4%Nb-2%Ta-0.2%Pd alloys were about 650 MPa and 800 MPa respectively, which were about 20~30% higher than those of the annealed alloys.

Keywords: titanium alloy, medical implant, microstructure, Vickers hardness, mechanical properties, corrosion fatigue strength, physiological saline solution, effect of heat treatment

1. Introduction

Stainless steel (SUS 316L), Co-Cr alloy, pure titanium and titanium alloys are being used as standard surgical implant materials. Among them Ti-6%Al-4%V ELI alloy is being viewed with keen interest, which originally had been used in aerospace applications. The cytotoxicity of pure metals and surgical implant materials have been reported by Steinemann⁽¹⁾. The high cytotoxicity of V and tissue response of capsule (scar tissue) type due to Al have been demonstrated, while Zr, Ti, Nb, Ta and Pt exhibited excellent biocompatibility and belong to loose connective vascularised (vital) group in the type of the tissue reaction. Also it is reported that the rate of colony formation (%) = (number of colonies in each concentration / number of control colonies) × 100 of V79 cells originating from the lung fibroblast in a Chinese hamster decreases linearly from V ion concentration of 0.16 mass ppm, and no colony is formed in a medium solution containing 1.1 mass ppm of

V ion⁽²⁾. As $\alpha + \beta$ type alloys for medical implants not containing V, alloys such as Ti-6%Al-7%Nb, Ti-5%Al-2.5%Fe, Ti-6%Al-6%Nb-1%Ta, Ti-5%Al-3%Mo-4%Zr, Ti-6%Al-2%Nb-1%Ta have been developed. As β type alloys, Ti-11.5%Mo-6%Zr-2%Fe alloy and Ti-15%Mo-5%Zr-3%Al⁽³⁾⁽⁴⁾, Ti-15%Mo-3%Nb-3%Al-0.2%Si, Ti-15%Mo-3%Nb-3%Al-0.3%O, Ti-11%Mo-6%Zr-4.5%Sn and Ti-13%Nb-13%Zr alloys⁽⁵⁾ have been developed for medical implant. A causal relationship between Al and the occurrence of senile dementia of Alzheimer type has been suggested by many researchers⁽⁶⁾⁻⁽¹²⁾. Moosmann et. al. reported that the concentration of alloying elements released from Ti-6%Al-4%V alloy to the peripheral tissues around the implant in the human body increases in the following order; Ti > Al > V⁽¹³⁾. Agins et. al. confirmed the presence of Ti, Al, and V in the patients who were failed cases of surgical implantation of Ti-6Al-4V alloy in hip joint. The histologic reaction that produced polynuclear leukocytes and giant cells was noted, and plasmacytes were frequently

observed if Al and V existed in large quantities in the tissue of the implanted part, causing the patient to present symptoms similar to allergic reaction and chronic inflammatory reaction⁽¹⁴⁾. Also, Kawahara reported that metallic Ti, Sn, Zr, Pd and Ta are low cytotoxic elements⁽¹⁵⁾. The toxicity of Sn and its compounds have been summarized by many researchers⁽¹⁶⁾⁻⁽¹⁸⁾ and it has been clarified that the toxicology of Sn is almost entirely the toxicity of organic compounds of Sn.

Thus, the pure metallic elements Zr, Nb, Ta, Pd and Sn, which can be expected to improve mechanical properties, corrosion resistance and biocompatibility could be selected as biocompatible alloying elements. Ti-15%Zr-4%Nb-4%Ta-0.2%Pd and Ti-15%Sn-4%Nb-2%Ta-0.2%Pd alloys without V and Al are currently being developed by the authors to excel in corrosion resistance, strength and ductility, and also to ensure long-term safety of use in the body for a period of more than 20 years as medical implants⁽²⁾⁽⁴⁾⁽¹⁹⁾⁻⁽²¹⁾. In this research, the effect of heat treatment on the mechanical properties at room temperature and the corrosion fatigue strength in a physiological saline solution of these Ti alloys were investigated.

II. Experimental method

Ti-15%Zr-4%Nb-4%Ta-0.2%Pd (Ti-Zr) and Ti-15%Sn-4%Nb-2%Ta-0.2%Pd (Ti-Sn) alloys designed with biocompatible alloying elements were melted by plasma electron beam. About 5 kg ingots were homogenized at 1273 ~1373 K for 7.2 ks, and then, they were hot-forged into round bars of 17 mm in diameter by β and α - β forgings. α - β forgings were carried out by reheating at 1123 K. Specimens of 17 mm in diameter and 10 mm in thickness cut from the sample alloys were solution-treated in the temperature range of 953~1173 K for 3.6 ks at intervals of 20 K, and then they were water-quenched. Hot-rolled block specimens with a thickness of 5 mm were also solution-treated under the optimized solution treatment conditions which included 15~20 vol% of primary α phase content (1038 K for 3.6 ks for Ti-Zr alloy, 1098 K for 3.6 ks for Ti-Sn alloy). The block specimens of 10 × 10 × 5 mm were cut from the sample alloys, aged in the temperature range of 623~873 K for 0.6~108 ks (10 min~30 h) at intervals of 50 K, which were then air-cooled. For the purpose of comparison, specimens were also annealed at 973 K for 7.2 ks.

Moreover pure Ti grade 2 and Ti-6%Al-4%V ELI alloy were melted by vacuum arc melting. After β and α - β forging, the alloys were annealed for 7.2 ks at 973 K. β forging : 1373 K, α - β forging : 1173 K ~1223 K (Ti-6%Al-4%V ELI) and 1103 K ~1123 K (pure Ti grade 2). Table 1 shows the chemical composition of these materials.

After polishing with waterproof emery paper of #120~#4000 and buff-cleaning with OP-S suspension, Vickers hardness was measured. Also after polishing and buff-cleaning, the test specimens were etched in hydrofluoric nitrate solution containing 10% H₂O₂ (HF: HNO₃: H₂O=15cm³:20 cm³:65cm³). The microstructure was observed by using an optical microscope, a scanning electron microscope (SEM) and a transmission electron microscope (TEM-2000EX), and the phases were identified with a X-ray diffractometer. The sample specimen for transmission electron microscope observation was prepared by electrolytic polishing (35 V, 50 to 70mA) with 95% acetic anhydride+5% perchloric acid solution. X-ray diffraction profiles were obtained using Fe K α X-ray at a tube voltage of 40 kV, tube current of 0.03 A and scanning rate (2 θ) of 1° / min. Moreover, to examine the room temperature strengths of the alloys after annealing, solution treatment and aging treatment, tensile testings were conducted with test specimens, 6 mm in diameter and 22 mm in gage length with a crosshead speed of 8.33 × 10⁻⁶ m/s (0.5mm/min).

The fatigue test specimen with dimension and shape as shown in Figure 1 was cut from the sample alloys. The corrosion fatigue test was carried out in a physiological saline solution containing 10% sodium bicarbonate solution (20 ml) and fetal bovine serum (5 ml) added to Eagle's MEM solution (H₂O:100 ml, eagle's MEM:0.94 g). To remove the residual strain on the surface of alloy caused in the manufacturing process, the surface of the test specimen was fully ground with #600 waterproof emery paper in the direction parallel to the test specimen. The test specimen was fitted inside a cell (polyethylene, inner diameter : 40 mm, outer diameter : 50 mm, height : 55 mm) containing Eagle's MEM+fetal bovine serum solution and then the test specimen was set on a corrosion fatigue testing machine (Servopulsar; manufactured by Shimadzu Co.). The corrosion fatigue test was conducted by circulating hot water at about 42 °C around the cell containing the physiological saline solution so as to maintain 310 K (37 °C) inside the cell. To make the oxygen concentration in the saline solution closer to that of the living body, and to maintain pH value at 7.4,

Table 1 Chemical composition (mass%) of materials used .

Titanium alloy	Zr	Sn	Nb	Ta	Pd	Al	V	Mo	Fe	O	N	H	C	Ti
Pure Ti grade 2	—	—	—	—	—	—	—	—	0.062	0.078	0.0037	0.0012	0.007	Bal.
Ti - 6%Al - 4%V ELI	—	—	—	—	—	6.42	4.19	—	0.198	0.101	—	0.0052	0.011	Bal.
Ti - 15%Zr - 4%Nb - 4%Ta - 0.2%Pd	15.1	—	3.97	3.86	0.18	—	—	—	—	0.096	0.0046	0.0013	—	Bal.
Ti - 15%Sn - 4%Nb - 2%Ta - 0.2%Pd	—	14.74	4.12	2.02	0.18	—	—	—	—	0.054	0.0038	0.0011	—	Bal.

a small amount of 90%N₂ + 5%CO₂ + 5%O₂ mixed gas was bubbled through the solution. As the corrosion test conditions, a sine wave having an R ratio (minimum tensile stress)/(maximum tensile stress) of 0.1, a frequency of 10 Hz, and a number of cycles up to 10⁸ times were maintained.

Fig.1 Dimensions of test specimen for corrosion fatigue.

III. Results and Discussion

1. Vickers Hardness and Microstructure of Solution-treated Alloys

Figure 2 shows the effect of solution treating temperature on Vickers hardness and microstructure. The Vickers hardness of Ti-Zr alloy increased linearly with increasing solution treating temperature up to 953~1033 K (680~760 °C), but remained almost constant at temperatures greater than 1053 K (780 °C). The Vickers hardness of Ti-Sn alloy showed a tendency to decrease with increasing solution treating temperature up to 953~1053 K (680~780 °C), but otherwise increased with increasing solution treating temperature in the temperature range of 1053~1113 K (780~840 °C), and thereafter remained nearly constant. At the solution treating temperature of 973 K, Ti-Zr alloy showed a two-phase (α + β) microstructure, and the β phase grew with increasing solution treating temperature, and at 1033 K (760 °C), the α phase existed as islands in the β phase matrix. The β phase appears white in the SEM micrographs, but X-ray diffraction profiles indicated that it was almost transformed to α' martensite by water quenching. The single β phase maintained at 1053 K (780 °C) was transformed to α' martensite during water quenching. The Ti-Sn alloy showed a similar tendency, except that, the β transition temperature was high. These observations were in conformation with the β transus obtained by differential thermal analysis (1055 K for Ti-Sn alloy, 1119 K for Ti-Sn alloy). However, the X-ray diffraction profiles of the Ti-Sn alloy indicated the peak of α' martensite.

In order to determine the optimum solution treating temperature, the volume fraction of primary α phase was examined. As shown in Figure 3, the volume fraction of primary α phase decreased markedly with increasing solution treating temperature in both Ti-Zr and Ti-Sn alloys. Optimized solution treatment conditions containing 15~25% volume fraction of primary α phase were selected at 1038 K (765 °C)-3.6 ks for Ti-Zr alloy and 1098 K (825 °C)-3.6 ks for Ti-Sn alloy.

Fig.2 Effect of solution treating temperature on Vickers hardness and microstructure.

Fig.3 Change in volume fraction of primary α phase with solution treating temperature

2. Hardness and Microstructure of Alloy Aged After Solution Treating

Figure 4 shows the changes in Vickers hardness of Ti-Zr and Ti-Sn alloys with aging temperature and aging time. The Vickers hardness of both alloys increased markedly for 0.6 ks aging at 673 K and 723 K. It was maximum after aging for 36 ks, then showed a gradual downward trend. Vickers hardness of Ti-Zr alloy increased for 0.6 ks aging at 773 K, but declined with the increase of aging time, and went down to a value lower than that of the solution-treated alloy at temperatures above 823 K (550 °C). The Ti-Sn alloy showed a similar tendency, with its Vickers hardness decreasing with increasing aging temperature. Figure 5 shows SEM micrographs of solution-treated and aged Ti-Sn alloy. In the case of the solution-treated alloy shown in Figure 5(a), the primary α phase and β phase constitute the transformed α' martensitic matrix. The solution-treated alloy showed no change in the SEM micrographs even after aging at 723 K for 36 ks.

Fig.4 Change in Vickers hardness with aging temperature and aging time. (a): Ti-15%Zr-4%Nb-4%Ta-0.2%Pd alloy, (b): Ti-15%Sn-4%Nb-2%Ta-0.2%Pd alloy.

However, aging at 823 K (550 °C) caused the β phase precipitation in the primary α region, and β phase precipitates observed in the matrix increased with increasing aging temperature. Figure 6 shows the X-ray diffraction profiles of solution-treated and aged Ti-Sn alloy. As seen in the figure, primary α and α' martensite as well as the peak of α' martensite were observed for the solution-treated alloy, and this α' martensite peak disappeared when the alloy was aged at 723 K (450 °C) for 36 ks. As for the alloy aged at 873 K for 36 ks, the β phase peak generated by aging was observed. Since a similar tendency was observed with Ti-Zr alloy, it is considered that phase precipitation due to aging causes a decline in Vickers hardness. Optimized aging treatment conditions for securing the maximum hardness were determined at 673 K -36 ks for Ti-Zr alloy and 723 K -36 ks for Ti-Sn alloy.

3. Mechanical Properties of Solution-treated and Aged Alloys

Table 2 shows the room temperature strengths of Ti-Zr and Ti-Sn alloys solution-treated and aged under optimized conditions compared to those of annealing. The 0.2% proof strength and tensile strength of both alloys increased after solution treatment and aging, but the elongation and reduction of area of the solution-treated and aged alloys were lower than those of annealed alloys. The strength and ductility of both alloys were higher than the ASTM values specified for Ti-6%Al-4%V ELI for surgical implantation (F136 -84)⁽²²⁾.

Fig.5 SEM micrographs of Ti-15%Sn-4%Nb-2%Ta-0.2%Pd alloy. (a):Solution treatment (S.T.), (b):S.T.+ aging at 723 K for 36 ks, (c):S.T.+ aging at 773 K for 36 ks, (d): S.T.+ aging at 873 K for 36 ks.

Fig.6 X-ray diffraction profiles of Ti-15%Sn-4%Nb-2%Ta-0.2%Pd. (a): Solution treatment (S.T.), (b): S.T.+aging at 723 K for 36 ks, (c): S.T.+ aging at 873 K for 36 ks.

When the new alloy is used as the stems of artificial hip joints, its Young's modulus should be as close to that of compact bone as possible because the greater the difference between the two values, the more the patient becomes vulnerable to the loosening of bone interface and bone fracture. While the Young's modulus of bone is as low as about 20 GPa, Ti-6%Al-4%V ELI and the new Ti alloy for surgical implantation have nearly the same modulus of 100 GPa and 90~100 GPa, respectively, which is comparatively closest to the value of compact bone as compared to SUS316L (200 GPa) and Co-Cr (200 ~ 230 GPa) alloys.

Table 2 Comparison of mechanical properties of alloys annealed and aged after solution treatment with those of Ti-6%Al-4%V ELI.

Mechanical properties	Alloys	Ti - 6%Al - 4%V ELI	Ti - 15%Zr - 4%Ta - 0.2%Pd		Ti - 15%Sn - 4%Nb - 0.2%Ta - 0.2%Pd	
		ASTM	Annealing	S.T. + Aging	Annealing	S.T. + Aging
0.2%Proof strength $\sigma_{0.2\%Pd}$ / MPa		795	693	806	790	1020
Tensile strength σ_{UTS} / MPa		860	715	919	860	1109
Total elongation %		10	28	18	21	10
Reduction of area %		25	67	72	64	39
Young's modulus E / GPa		100	94	99	89	103

S.T. : Solution treatment

4. TEM Micrographs of Solution-treated and Aged Alloys

Microstructures of the test alloys were examined using a transmission electron microscope in order to clarify the reasons for the increase in their hardness and strength after solution treatment and aging. Figure 7 shows SEM and TEM micrographs of the alloys after annealing at 973 K for 7.2 ks. The white phase noted in the SEM micrographs is β phase at 973 K which is transformed almost entirely to α phase during cooling, as is clear from the X-ray diffraction profiles shown in Figure 8. The TEM micrographs in Figure 7(b) show the cell structure of α phase transformed from β phase. The β phase seen in the X-ray diffraction profiles of Ti-Sn alloy envelops the α phase and appears black in the TEM micrographs as shown in Figure 7(e).

Fig.7 Optical, SEM and TEM micrographs after annealing at 973 K for 7.2 ks. (a) and (c): Optical, and SEM micrographs, (b), (d), (f): TEM micrographs (a) and (b): Ti-Zr alloy, (c)-(e): Ti-Sn alloy.

Although the results are different from the chemical analysis shown in Table 1, quantitative analysis using EDX disclosed that β phase (Sn: 10.7 mass%, Nb: 14.7%, Ta: 2.8%) contained a little less Sn and a little more Nb and Ta than the α phase matrix (Sn: 12.8%, Nb: 4.3%, Ta: small amount). Nb and Ta are β stabilizing elements while Sn is considered to be α stabilizing element.

Fig.8 X-ray diffraction profiles after annealing at 973 K for 7.2 ks.

Figure 9 shows the TEM micrographs of Ti-Zr alloy aged after solution treatment. As seen in the figure, primary α and lath-like α' martensite can be observed for the solution-treated alloy, while α phase precipitation at intersections of lath-like martensite and the growth of α phase with increasing aging time can be observed for the alloy aged at 673 K for 1.8 ks. It can also be seen that recrystallization and growth of α phase precipitates in the lath-like α' martensitic matrix increases with increasing aging temperature. Figure 10 shows the TEM micrographs of Ti-Sn alloy. Although the martensite of the solution-treated alloy shows certain differences, fine α phase precipitates due to aging can be observed similar to those of Ti-Zr alloy. The concentration of alloying elements in the primary α phase and martensite after aging was examined using EDX, and the results are shown in Table 3. The concentration of alloying elements in the new α phase precipitation by aging nearly coincided with that in primary α phase after solution treatments. The concentration of β phase stabilizing elements, namely, Nb and Ta, tend to be higher in α' martensite than in α phase.

Fig.9 TEM micrographs of Ti-15%Zr-4%Nb-4%Ta-0.2%Pd alloy aged at 673~873 K for 1.8~36 ks after solution treatment at 1098 K for 3.6 ks. (a): Solution treatment (S.T.), (b): S.T.+aging at 673 K for 1.8 ks, (c): S.T.+ aging at 673 K for 36 ks, (d): S.T.+aging at 773 K for 36 ks, (e): S.T.+ aging at 873 K for 36 ks.

Table 3 Comparison of concentration (mass%) of alloying elements in primary α phase after solution treatment and α phase precipitated in aging.

	Ti- Zr				Ti- Sn		
		α	α'			α	α'
S.T.	Zr	13.2	18.9	S.T.	Sn	13.6	11.7
	Nb	2.5	5.8		Nb	3.0	7.7
	Ta	2.4	3.2		Ta	—	1.5
S.T. + 673K - 1.8ks	Zr	13.9	17.6	S.T. + 673K - 1.8ks	Sn	13.8	13.3
	Nb	1.8	5.0		Nb	3.3	6.4
	Ta	2.2	2.7		Ta	—	1.1
S.T. + 873K - 36ks	Zr	17.4	19.7	S.T. + 873K - 36ks	Sn	12.7	13.2
	Nb	2.6	6.4		Nb	2.8	6.6
	Ta	2.5	4.6		Ta	0.8	1.1

As described above, it was revealed that when the alloy is solution-treated immediately below the β transition temperature and aged at 673~723 K for 36 ks, α phase precipitates from the lath intersections of the fine α' martensite produced by solution treating, and that the alloy strength can be increased to maximum extent when the α phase precipitates are finely and uniformly distributed.

5. Corrosion Fatigue Strength

When the current density measured by anodic polarization testing in a physiological saline solutions becomes too low, the corrosion resistance of implants in the living body must be excellent⁽³⁾⁽²⁰⁾.

Fig.10 TEM micrographs of Ti-15%Sn-4%Nb-2%Ta-0.2%Pd alloy aged at 623~873 K for 1.8~36 ks after solution treatment at 1098 K for 3.6 ks. (a): Solution treatment (S.T.), (b): S.T.+aging at 623 K for 36 ks, (c): S.T.+ aging at 723 K for 1.8 ks, (d): S.T.+aging at 723 K for 36 ks, (e): S.T. + aging at 823 K for 36 ks, (f): S.T.+ aging at 873 K for 36 ks.

Figure 11 shows anodic polarization curves measured in 450 ml Eagle's MEM+50 ml fetal bovine serum solution deaerated with high-purity N_2 gas bubbling at 310 K (37 °C). The current density in the electric potential region of 1 V or more increases as the electric potential increases and increases rapidly in the potential region of 2 V or more. Current density at higher voltage region more than 2 volts tend to be higher in the following relation ; Ti-6%Al-4%V ELI > Pure Ti grade 2 > Ti-15% Sn-4%Nb-2%Ta-0.2%Pd > Ti-15%Zr-4%Nb-4%Ta-0.2% Pd alloy⁽¹⁷⁾.

Fig.11 Comparison of the anodic polarization curves for medical implants in a deaerated Eagle's MEM + bovine serum at 310 K.

Figure 12 shows the S-N curves obtained from the corrosion fatigue test conducted in the physiological saline solution. Comparison of Ti-15%Sn-4%Nb-2%Ta-0.2%Pd with Ti-15%Zr-4%Nb-4%Ta-0.2%Pd indicates that Ti-Sn alloys having a higher room temperature strength tend to have a larger number of cycles to failure in both case of annealing and solution treatment+aging. The fatigue strength at 10^8 cycles in solution-treated+aged Ti-15%Zr-4%Nb-4%Ta-0.2%Pd and Ti-15%Sn-4%Nb-4%Ta-0.2%Pd alloys were about 650 MPa and 800 MPa respectively, and they were about 20~30% higher than those of the annealed alloys.

Fig.12 Effect of heat treatment on the S-N curves obtained by corrosion fatigue test in a physiological saline solution at 310 K.

IV. Conclusion

For the purpose of increasing the strength of Ti alloys for surgical implantation, they were solution-treated in the vicinity of β transition temperature and aged in the low temperature region, and the following results were obtained.

(1) Vickers hardness of both Ti-Zr and Ti-Sn alloys increases with increasing solution treating temperatures, and remains almost constant at temperatures higher than the β transition temperature. The alloy has a two-phase ($\alpha + \beta$) microstructure when the solution treating temperature is low, but the β phase grows with increasing solution treating temperatures, and a fine primary α phase precipitates in the β matrix immediately below the β transition temperature. Optimized solution treatment conditions for securing 15~20% volume fraction of primary α phase are 1038 K - 3.6 ks for Ti-Zr alloy and 1098 K - 3.6 ks for Ti-Sn alloy.

(2) By aging Ti-Zr alloy at 673 K for 36 ks and Ti-Sn alloy at 723 K for 36 ks after solution treatment, their strength can be increased while maintaining their ductility at a level equivalent to or higher than the ASTM value specified for Ti-6%Al-4%V ELI for surgical implantation. As far as the strength augmentation mechanism is concerned, it has been found that α phase precipitates from the lath intersections of fine α' martensite produced by solution treatment, and that the alloy strength is increased when the α phase precipitates are finely distributed.

(3) Comparison of corrosion fatigue strength of Ti-15%Sn-4%Nb-2%Ta-0.2%Pd and Ti-15%Zr-4%Nb-4%Ta-0.2%Pd alloy indicates that Ti-Sn alloy having a higher room temperature strength tend to have a larger number of cycles to failure in both annealing and solution treatment+aging. The fatigue strengths at 10^8 cycles in solution-treated and aged Ti-Zr and Ti-Sn alloys were about 650 MPa and 800 MPa respectively. These values are about 20~30% higher than those of annealed alloys.

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