REFERENCES

1. A. A. Goldenberg. Metalloved i Termich. Obrab. Met., 1959, No. 10, pp. 51-61. [HB Translation No. 4990].
2. H.W. Schleicher and U. Zwicker. Metallkunde, 1956, Vol. 47, No. 8, pp. 570-576.
3. H.M. Burte et al. Metal Progress, 1955, Vol. 67, No. 5, pp. 115-120.
4. A.D. McQuillan. Journal Institute of Metals, 1951, Vol. 79.

STRUCTURE AND MECHANICAL PROPERTIES OF HIGHLY ALLOYED TITANIUM

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This paper is concerned with the causes of the change of the mechanical properties of two medium-alloy titanium compositions, Table 1, after heat treatment. These alloys are used for forging billets and hot-rolled tubing. Depending on the heat treatment, they may have a strength of 25, 30 and more kg/sq.mm. These high values are achieved by quenching and aging. Heat treatment of thin-walled tubing and small section bars is limited to an aging treatment (without preliminary solutionizing).

TABLE 1

Alloy	Alloy content, %			
	Fe	Mn	Cr	Al
1	3	3	3	—
2	3	—	5	3

The structure of the alloys after forging, piercing and rolling contains three phases: β, α, ω. The amount of ω-phase is small and is not always detected with X-rays.

Under certain conditions, a eutectoid transformation is observed in these alloys as well as the formation of the metastable phases β and ω. Slugs for our investigation were prepared by sintering of titanium powder containing (%) 0.12 Fe, 0.074 Si, 0.12 Ni, 0.022 Al, 0.052.Ca, 0.003 H, 0.18 O, and 0.01 N. The sintered slugs were forged at 1000-700°C into 16 mm dia rods.

Inasmuch as the binary titanium alloys Ti-Fe, Ti-Cr, and Ti-Mn belong to the eutectoid system [1, 2], it was of interest to determine the tendency of these alloys to embrittle after holding for 100 hrs at 400, 500 and 600°C. Table 2 shows the mechanical properties of rods forged and annealed at 700 and 800°C, reheated as indicated above. Table 2 suggests that as the heating temperature is increased, the ductility of alloy 2 falls and that of alloy 1 increases.

In alloy 1, this change is associated with an increase of the α-phase. Heating of alloy 1, however, results in a transformation $\beta + \alpha + \omega \rightarrow \beta + \alpha$.

The strength of annealed alloy 1 after 100 hrs soaking at 400-600°C changes little; the increased ductilities are associated with a larger amount of α-phase. Fig. 1

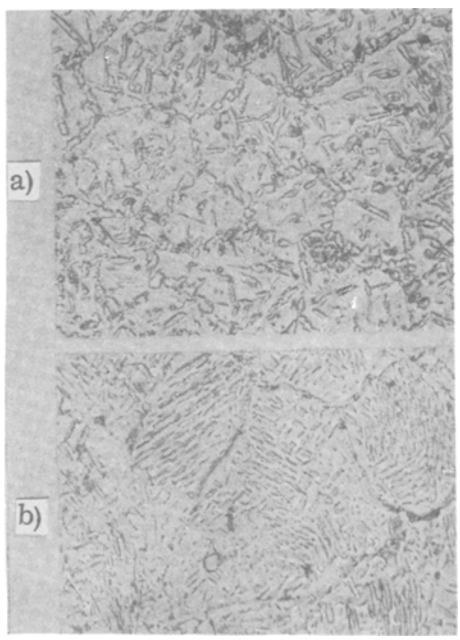

TABLE 2

Alloy no.	Heat treatment		Mechanical properties				
			R_C	σ_b	σ_s	δ	ψ
				kg/sq. mm		%	
1	Forged	. . . .	44	163,6	157,6	0,4	2,4
		400°, 100 hrs	46	167,0	163,4	4,2	6,1
		500°, 100 hrs	40	130,6	129,2	8,0	12,4
		600°, 100 hrs	38	122,3	120,4	19,2	26,3
	Annealed	700°, 1 hrs	37,5	129,1	126,1	12,0	17,0
		400°, 100 hrs	39	129,0	126,5	9,2	11,3
		500°, 100 hrs	38	130,3	129,7	17,0	31,6
		600°, 100 hrs	37,5	127,6	126,2	2,0	1,4
	Annealed	800°. 1 hrs	36,5	131,7	128,8	7,2	10,1
		400°, 100 hrs	38,5	132,0	127,7	15,2	21,8
		500°, 100 hrs	37,5	128,0	121,5	16,2	23,3
		600°, 100 hrs	37	125,1	120,0	10,0	15,1
2	Forged	. . . .	43	139,7	137,1	9,0	14,8
		400°. 100	44	170,0	166,1	1,2	2,0
		500°, 100	42	143,4	142,1	1,0	4,3
		600°, 100	41	143,7	131,0	3,2	7,0
	Annealed	800°, 1	38,5	121,2	118,4	15,8	20,1
		400°, 100	41	Ruptured at shoulder			
		500°, 100	39,5	133,5	—	2,0	11,2
		600°, 100	39	130,3	—	4,4	6,6

Note: Table contains average data of five tests.

shows the microstructure of alloy 1 after annealing at 800°C/1 hr and also after annealing at 800° and aging at 500°C/100 hrs. The increase of α content after aging is plainly seen.

The precipitation of α after heating at 400 and 500°C/100 hrs is due to the following factors: annealing at 700 and 800°C followed by cooling at about 100°C/hr does not produce an equilibrium condition. The content of β in the microstructure, Fig. 1, a, by far exceeds the equilibrium conditions. On aging at 400 and 500°C/100 hrs, a precipitation of α and an enrichment of the β -phase with β-stabilizers take place. The 100 hrs treatment does not bring the β -phase to the eutectoid concentration.

At 600°C a eutectoid concentration is reached after less than 100 hrs. When this happens, the following transformation takes place: $\beta \rightarrow \alpha$ + intermetallic compound. This change lowers the ductility, as follows from Table 2 (see mechanical properties of alloy 1 annealed at 700 and 800°C) after additional heating at 600°C/100 hrs.

The annealing temperature (700, 800°C) of alloy 1 affects the properties after long soaking. As a result of annealing at 700°C, the subsequent eutectoid decomposition at 600°C is accelerated and causes a more pronounced fall of ductility than annealing at 800°C. This effect depends on the composition of the metastable β -phase: the closer it is to the eutectoid concentration, the faster will the eutectoid transformation proceed.

The strength of forged alloy 2 after aging at 400°C/100 hrs exceeds the as-forged strength by 30 kg/sq. mm. Apparently a $\beta \rightarrow \omega$ transformation takes place here [3, 4]. However, we were unable to discover the ω-phase in this case by metallographic or X-ray methods. The extraordinary brittleness of annealed alloy 2 held at 400°C/100 hrs is apparently explained by the embrittling eutectoid transformation. This change of alloy 2 (both forged and annealed) is observed during holding at 500 and 600°C/100 hrs.

It is clear from these data that despite its 3% Al, alloy 2 has a stronger tendency to an embrittling eutectoid transformation than alloy 1. Apparently manganese exerts a hampering action on the rate of the eutectoid transformation in alloy 1. The curves in Fig. 2 show that alloys 1 and 2 contain a β-stabilizer above the critical value [4, 5].

Upon water quenching, the relatively soft β-phase is stablized: its hardness is 40 R_C in alloy 1 and 36.5 R_C in alloy 2.

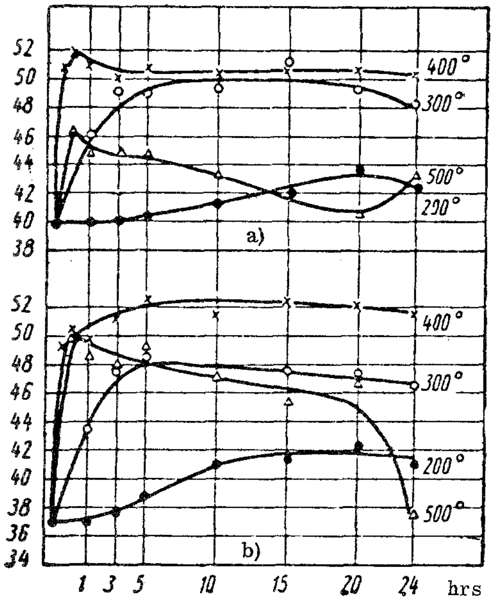

Fig. 2. Aging curves of alloys 1 and 2 after quenching from 900°C in water: a - alloy 1, b - alloy 2.

Aging at 200°C gradually hardens both alloys. After holding for 15-20 hrs, the hardness reaches a maximum and a further holding (up to 24 hrs) has practically no effect. Aging at 300° causes a hardness increase during the first 5 hrs but longer soaking adds very little.

At 400°C, a high saturation hardness (R_C 50) is reached during the first 15 to 30 min. The aging curve at 500°C shows a maximum hardness after about 15 to 30 min (1 hr max) followed by a decrease.

At 220-475°C a contraction of the specimen caused by ω precipitation is observed [4, 6]. Beginning from 475°C and above, the specimen expands due to vanishing of the ω-phase and appearance of $\alpha : \beta + \omega \rightarrow \beta + \alpha$.

Precipitation of α results simultaneously in a reduced hardness and improved ductility. Microstructural observations confirm the presence of an acicular α-phase after aging at 500°C/100 hrs. Hence, if at aging temperatures of 200, 300 and 400°C, only a precipitation of ω-phase is observed (after 24 hrs at 200 and 300°C the $\beta \rightarrow \omega$ transformation does not run to completion), at 500° first the ω-phase shows up (hardness increases) whereupon, after longer holding periods, the reaction $\beta + \omega \rightarrow \beta + \alpha$ follows and the hardness decreases. In the latter case, no eutectoid transformation was found.

REFERENCES

1. S.G. Glazunov and E.K. Molchanova. Phase Diagrams of Titanium Alloys (Diagrammy Sostoyanii Splavov Titana), [Book], Oborongiz, 1954.

2. G.V. Samsonov, V.S. Neshpor and L.V. Lange. Metalloved i Obrav. Met., 1956, No. 1, pp. 35-39.

3. Yu.A. Bagaryatskii, G.I. Nosova and T.V. Tagunova. Doklady Akad. Nauk SSSR, 1955, Vol. 105, No. 6, pp. 1225-1228. [HB Translation No. 4103].

4. Yu.A. Bagaryatskii, T.V. Tagunova and G.I. Nosova. Problems of Metal Science and Metal Physics (Problemy Metalloved i Fiz. Metallov), 1958, Vol. 5.

5. P.D. Frost and W.M. Parris. Transact. ASM, 1954, Vol. 46, pp. 231, 257 and 1056.

6. H. Böhm and H. Westphal. Zeitschrift Für Metalkunde, 1956, Vol. 47, pp. 558-563.

MECHANICAL AND TECHNOLOGICAL PROPERTIES OF TERNARY TITANIUM ALLOYS

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The paper of reference [1] deals with ternary titanium alloys Ti-Al-Mn, Ti-Al-Cr, Ti-Al-Mo, Ti-Al-Fe and Ti-Fe-Mn. Not counting the last, all others are based on the binary system Ti-Al with addition of one of the four β-stablizers.

The present work is based on the system Ti-Sn to which Zr, Cr, V, Mo and Mn were added. The system Ti-Al-Zr was also studied. In all these ternary systems we studied alloys along the section with 94% Ti and from 6% Sn to 6% of one of the β-stablizers listed.

Preparation and Testing of Alloys. The alloys were prepared from pre-mixed sponge of one batch. The strength of sheet made of this material was 55.5 kg/sq. mm; elongation, 32.7% (0.3% Fe, 0.15% Si). The sponge was alloyed with manganese, refined chromium, aluminum, vanadium, tin (metallic), iodide zirconium, technically pure iron and molybdenum powder.

The ingots were prepared in a vacuum arc furnace with a double ('stepped') mold using a consumable electrode. The weight of the charge was 3 kg. The ingots were forged and rolled under laboratory conditions according to the existing technology. The sand blasted and pickled sheets 1.3-1.5 mm thick were vacuum annealed (5×10^{-3} mm Hg) at 800°C/2 hrs, furnace-cooled to 200°C and then cooled in air. The composition was determined on the finished product, Table 1.