STRUCTURE, TEXTURE, AND PROPERTIES OF ALLOY VT20 AFTER DEFORMATION AND ANNEALING BY DIFFERENT SCHEDULES

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The structure, texture, and mechanical properties of four melts of alloy VT20 (Table 1) after rolling and heat treatment have been studied.

Alloy structure was evaluated in a MIM-8 metallurgical microscope; phase composition and texture were evaluated in a DRON-3 x-ray diffractometer. Limiting deformability was deter-

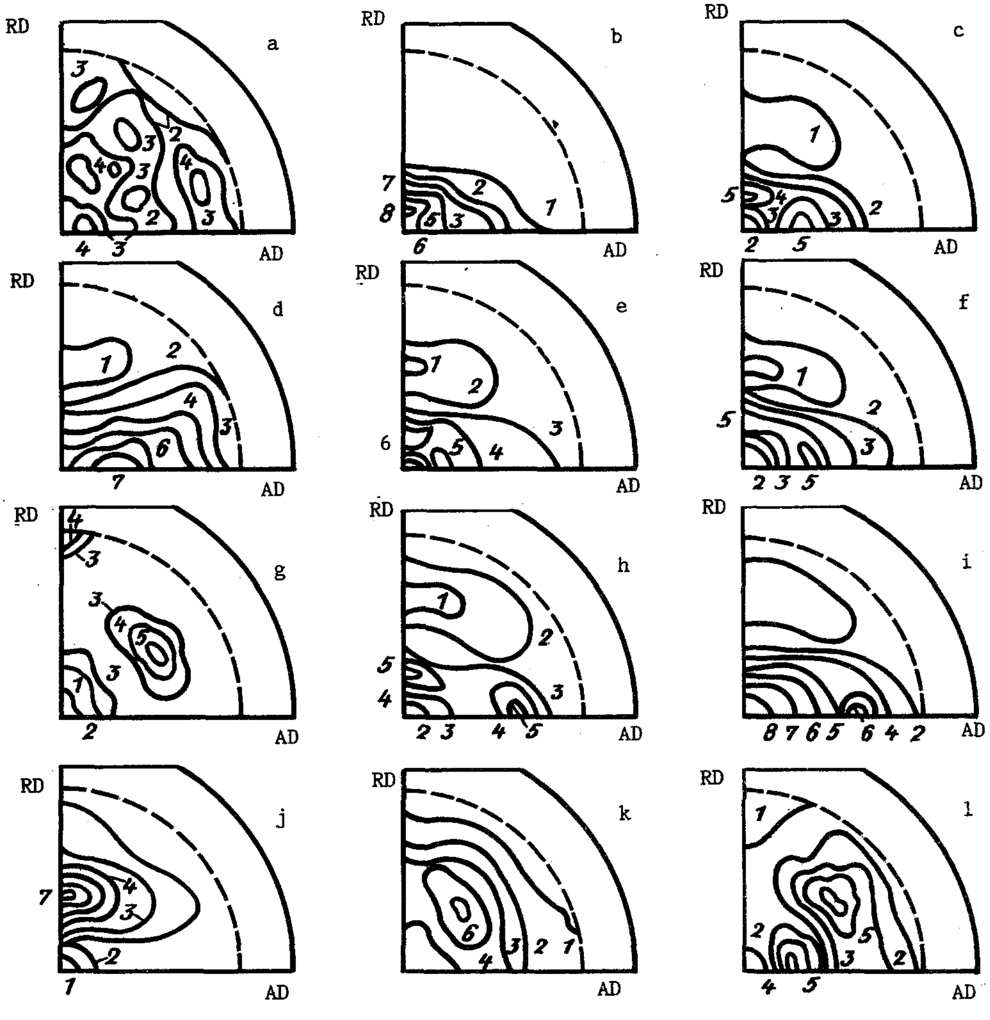

Fig. 1. Polar diagrams (0002) for alloy VT20: a) blank; b) melt 1; c, e-i) 2; d) 3; j-l) 4; b-d) original condition; e, f) reduction ε = 30 and 80% respectively; g) same as e, plane {10$\bar{1}$0}; h, i) deformation with ε = 80% and vacuum annealing for 1 h at 670 and 770°C respectively; j, k) outer and inner layer of specimens after heat treatment by schedule 6; l) treatment by schedule 7.

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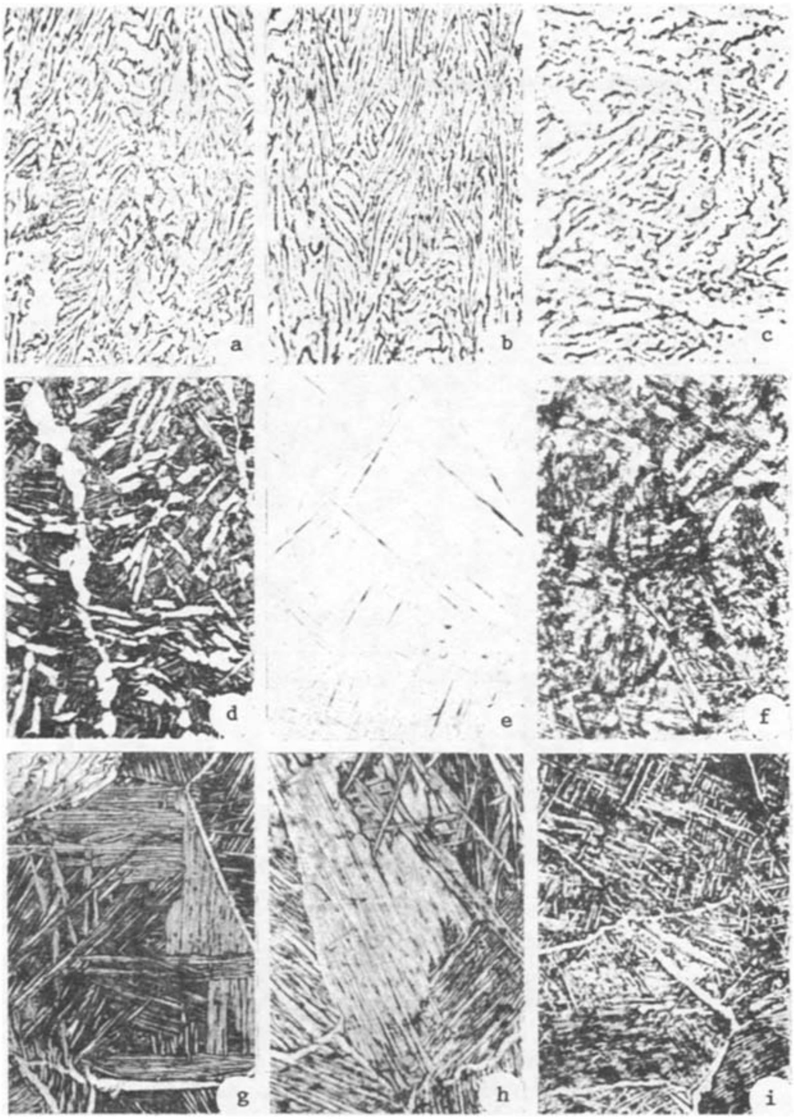

Fig. 2. Microstructure of alloy VT20 specimens (×500): a) cold rolling with reduction of 10%; b) 30%; c) heat treatment by schedule 1; d) 2; e) 3; f) 4; g) 5; h) 6; i) 7.

TABLE 1

Melt	Content of alloying elements, %			
	Al	Zr	V	Mo
1	6,7	2,6	1,4	1,3
2	6,9	2,1	1,7	1,2
3	7,0	2,8	1,5	1,4
4	7,1	1,8	1,4	1,6

Note. Apart from the elements listed the melts contained not more than 0.05% C; 0.1% Fe, 0.07% Si; 0.015% H; 0.02% N; 0.09% O.

mined by rolling wedge-shaped specimens. Mechanical properties of specimens were determined in tensile tests.

X-ray studies showed that blanks of alloy VT20 contain from 3 to 5% β-phase, and their structure is characterized by the presence of individual areas of increased polar density (Fig. 1a). Absence of symmetry in polar diagrams (PD) points to the existence of texture in-

TABLE 2

Heat treatment schedule	β-phase %	σ_f	$\sigma_{0,2}$	ψ	δ
		N/mm²		%	
Anneling at 790°C for 8 h in a vacuum (in the furnace) + annealing at 790°C for 0.5 h in a vacuum (air) (1)	8	1000	910	25,0	10,0
Annealing at 790°C for 8 h in a vacuum (air) + annealing at 1010°C for 0.25 h (air) + 730°C for 0.25 h (air) (2)	6	1040	950	28,0	11,5
Heating to 1100°C (water) (3)	2	1060	960	24,0	9,0
Heating to 1100°C (air) (4)	10	1030	950	26,5	10,5
Heating to 1100°C, v_{cool} = 5°C sec (5)	11	1040	960	27,0	11,0
Heating to 1100°C, v_{cool} = 2.5°C/sec (6)	12	1035	950	30,5	10,5
Schedule 4 + annealing at 800°C for 0.5 h (air) (7)	10	1020	940	28.8	10.0

Note. 1. Given in brackets is cooling medium. 2. Specimen heating in schedule 3-6 was carried out at a rate of 25°C/sec.

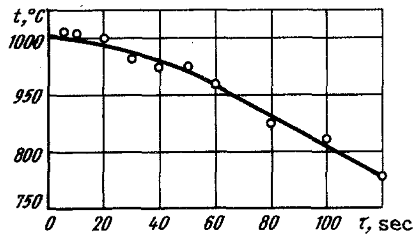

Fig. 3. Dependence of temperature for the end of recrystallization for alloy VT20 specimens deformed by cold rolling with a reduction of 40% with v_h = 25°C/sec on annealing duration (correction: along the ordinate axis instead of 750 and 800°C read 850 and 900°C respectively).

heritance for multivant β-phase transformation during cooling and cooling and conversion of blank material in the α-phase temperature region.

Shown in Fig. 2 is the microstructure of alloy VT20 after cold rolling and heat treatment by different schedules.

Cold rolling with low levels of reduction (apparoximately up to 10%) leads to development of slip bands (Fig. 2a) which points to the action of some slip systems in the deformation process. An increase in the degree of deformation to 30%, apart from the development of slip, leads to fragmentation of the structure (Fig. 2b). A further increase in production causes marked elongation of structural components along the rolling direction (RD). It should be noted that no marked contribution of twinning to alloy deformation was observed. This is connected with a marked content of aluminum impurity, which as is well known [1], makes it difficult to develop twinning in titanium.

The original alloy specimens prepared by hot and warm rolling of blanks exhibited three types of texture (see Fig. 1b-d): 1) maxima for polar density are diverted in the RD from a direction normal to the rolling plane (RP), which in terms of ideal orientations may be expressed as: (0001) ± β_1RP — AD; 2) maxima are diverted in the RD and in a direction across the RD (AD): (0001)± α_1RP — AD ± β_2RP — AD; 3) maxima are diverted in the AD; (0001) ± α_2RP — AD (here α and β are angles of deviation for polar density maxima from the RD in the AD and in the RD rsepectively).

Cold rolling of original alloy specimens causes a marked change in texture [2]. With reductions up to 25-30% in specimens with texture type 1 and 3 a two-component texture type 2 forms. An increase in reduction to 35-40% causes a sharp approach of the maxima for basal poles towards the RD and a marked increase in the intensity of the central part of PD (Fig. 1e), which leads to a reduction in specimen deformability as a result of making basal slip difficult and the disappearance of easy deformation systems. A reduction in deformability may cause material failure with reduction of the order of 30-40% as a result of forming in them an unfavorable crystallographic texture.

An increase in reduction by split deformations of the order of 2.5% for a pass causes an increase in the slope of basal planes towards the RD (Fig. 1f, g), which aids alloy deformability as a result of including new slip systems and it makes it possible to roll the alloy to high (80-90%) total reductions.

It is possible to increase alloy ductility by using interpass annealing in a protective gas atmosphere, vacuum, or in air; its duration should be reduced to the maximum degree and heating temperature should be reduced. In worked specimens the latter should be achieved as a result of a reduction in the recrystallization temperature t_r [3], which in specimens with a reduction of about 80% is 770°C. After annealing in a vacuum at this temperature there is formation of a recrystallized structure in alloy specimens (Fig. 1l).

It is possible to reduce exposure in air by using high-speed heating (electrical heating). With high-speed heating the dependence of recrystallization temperature on the degree of deformation is reduced, which is connected with suppression of recovery processes [4], and with an increase in heating rate the recrystallization process shifts into the high-temperature region (for the test alloy with v = 25°C/sec by 100°C compared with furnace heating). A reduction in recrystallization temperature with rapid heating is possible as a result of isothermal soaking (Fig. 3).

The ductility of titanium alloys containing β-phase is determined by the amount of it and stability [5, 6]. An increase in the content of stable β-phase promotes an increase in alloy ductility since it is more ductile than α-phase in which preferential dissolution of aluminum leads to intense strengthening of it. Metastable β-phase, which it is possible to assess from the difference between σ_f and $\sigma_{0.2}$ [7], during plastic deformation undergoes martensitic transformation and it is considerably strengthened. Thus, in order to improve alloy ductility heat treatment before cold rolling should provide the maximum amount of stable β-phase.

Heat treatment by schedules 1 and 2 (Table 2), including prolonged vacuum annealing, makes it possible to obtain a stable structure and a favorable combination of σ_f and $\sigma_{0.2}$ with quite a high level of ductility characteristics. Alloy microstructure after heat treatment by schedule 1 is a finely-dispersed mixture of α- and β-phases (see Fig. 2c), and after treatment by schedule 2 it is lamellar α-phase obtained as a result of β → α-transformation with disseminated islands of primary α-phase of almost polyhedral shape (Fig. 2d). The higher level of alloy ductility after treatment by schedule is apparently obtained as a result of partial phase recrystallization with heating to 1010°C.

By considering this favorable factor a study was made of the effect of high-speed heat treatment with heating to the β-region temperature (1100°C) which provides complete alloy phase recrystallization. A high heating rate suppresses intense grain growth above the α + β → β-transformation temperature. Alloy cooling was carried out in air, in water, and with rates of 5 and 2.5°C/sec. Some of the specimens cooled in air were given a stabilizing anneal at 800°C for 30 min followed by air cooling.

Specimens cooled in water exhibit low deformability, which is connected with presence in them of martensitic α'-phase and a small (approximately 2%) content of β-phase. The mechanical properties of these specimens are characterized by high strength and low ductility. The maximum content of β-phase in the alloy (about 12%) is recorded in specimens cooled at a rate of 2.5°C/sec; as cooling rate increases its amount decreases sharply. Heat treatment by schedules 4-6 provides sufficient alloy softening with a favorable difference between σ_f and $\sigma_{0.2}$.

Comparison of alloy microstructure after heat treatment by schedules 3-6 (Fig. 2e-h) points to a marked effect of cooling rate on grain size; the finest grain size is observed in specimens cooled in air (Fig. 2f), and the coarsest in specimens cooled at a rate of 2.5°C/sec (Fig. 2h). In addition, the microstructure of the latter is characterized by a coarse intragranular structure.

The advantage established from the point of view of β-phase stability and amount for cooling at low rates (2.5 and 5°C/sec) is difficult to realize under production conditions for which air cooling is practicable. In this case a favorable effect on the structure may be repeated annealing in the furnace. As can be seen from Fig. 2i and Table 2, additional annealing (schedule 7) has a weak effect on the mechanical properties and structure of specimens, although β-phase is markedly stabilized.

The effect of annealing on texture was studied in specimens after heat treatment in a vacuum furnace and an electrical contact device. Vacuum annealing in a furnace at temperatures above the recrsytallization temperature leads to formation of a recrystallized texture which is characterized by a considerable increase in the intensity of basal poles in the RD and an increase in their scattering in the AD (see Fig. 1i). In cold rolled specimens along the rolling direction there is predominantly orientation of crystallographic direction $[11\bar{2}0]$ instead of $[10\bar{1}0]$ (Fig. 1g). Heat treatment by schedule 6 leads to formation of type 1 texture with maxima in the rolling direction (Fig. 1k) which are at larger angles than in specimens of melt 1 (Fig. 1b). In the inner layers of specimens after electric heating maxima for basal planes occupy an intermediate position between the RD and AD (Fig. 1k). The difference in texture of inner and outer specimen layers after high-speed treatment may be explained by the difference in their cooling conditions after heating and formation of a texture with an intermediate position for maxima, i.e., development of polygonization and incomplete recrystallization processes [8, 9]. This also explains the form of the texture of internal layers subjected to additional annealing (Fig. 1l).

CONCLUSIONS

1. Cold rolling of alloy specimens leads to formation of a two-component texture which with deformation at the level of about 30% approaches the baseline. This creates a region of critical reduction (27-40%) were material failure is possible. This region may be passed by repeated rolling with small (approximately 2.5%) reductions per pass. With further deformation (50% or more) a texture of the deviation type forms which makes it possible to achieve (80-90%) reductions.

2. In order to obtain good structural, textural, and mechanical characteristics it is recommended that vacuum annealing in the furance or high-speed heating above the phase recrystallization temperature with cooling in air is carried out after cold rolling. Subsequent short-term annealing in the furnace at 800°C with air cooling was a favorable stabilizing effect on alloy structure.

LITERATURE CITED

1. B. A. Kolachev, Physical Metallography of Titanium [in Russian], Metallurgiya, Moscow (1976).
2. A. A. Bryukhanov, I. G. Zakhavchenko, and A. F. Tarasov, "Anisotropy of elastic properties and textural strengthening of alloy VT20 sheets," Fiz. Met. Metalloved., 54, No. 2, 415-416 (1982).
3. S. S. Gorelik, Recrystallization in Metals and Alloys [in Russian], Metallurgiya, Moscow (1978).
4. A. I. Gordienko and A. A. Shipko, Structural and Phase Transformations in Titanium Alloys with Rapid Heating [in Russian], Nauka i Tekhnika, Moscow (1983).
5. B. A. Kolachev, R. M. Gabidulin, and Yu. V. Piguzov, Heat Treatment Technology for Nonferrous Metals and Alloys [in Russian], Metallurgiya, Moscow (1980).
6. V. K. Aleksandrov, N. F. Anoshkin, G. A. Bochvar, et al., Titanium Alloy Semifinished Products [in Russian], Metallurgiya, Moscow (1979).
7. V. N. Gridnev, O. M. Ivasishin, and S. P. Oshkaderov, Physical Bases of Rapid Thermal Hardening of Titanium Alloys [in Russian], Naukova Dumka, Kiev (1986).
8. Yu. S. Bratchikov and R. A. Adamesku, "Structural changes in cold-rolled titanium in the polygonization stage," in: Heat Treatment and Physics of Metals: Collected Works, No. 2, Sverdlovsk (1976).
9. A. A. Bryukhanov and A. F. Tarasov, "Texture and anisotropy of the properties for some titanium and zirconium alloys with deformation and heat treatment," MS dep. in UkrNIITI 12.02.86, No. 534-Uk86.