PECULIARITIES OF PHASE AND STRUCTURE TRANSFORMATIONS IN TITANIUM ALLOYS IN THE PROCESS OF HIGH-SPEED ELECTRIC HEAT TREATMENT

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Introduction

In a number of papers (1-5) it was established that a high speed electric heat treatment together with plastic deformation may be considered one of the effective methods of structure refinement and improvement of titanium alloy mechanical properties. At the same time many problems concerning the mechanism and kinetics of phase transformations often proceeding under the conditions of high-speed continuous heating simultaneously with polygonization and recrystallization have been insufficiently investigated.

The study of phase transformation mechanism and kinetics, structure and properties was conducted using a Ti-6Al-1Mo-1V alloy having an initial coarse-grained lamellar structure and BT6, BT9 and BT3-1 commercial alloys characterized by a fine-grained equiaxial structure. The temperatures of complete polymeric transformation and the volume fraction of $\alpha$ phase in an annealed condition in the above alloys tested were 1020°C and 9%, 960°C and 10%, 980°C and 13% and 970°C and 17% respectively. In order to investigate the effects of conditions of furnace and high-speed continuous annealings on the process of recrystallization the blanks made of alloys having coarse-grained lamellar structure, $D=500$ to $1000\ \mu m$, $d$ (size of a lamella bundle)=$60$ to $120\ \mu m$, $b$ (lamella thickness)=$3$ to $5\ \mu m$ were 10-to-70% reduced at $900°C$ to $950°C$ at a rate of $10.2\ s^{-1}$ with a subsequent cooling in air.

The heating at rates between $1°C/s$ and $1500°C/s$ was performed on a apparatus of resistance heating by direct passing a current through a blank. The preset temperature being achieved the specimens were cooled in water.

Mechanism and kinetics of $\alpha + \beta \rightarrow \beta$ Phase Transformation

The results of experiments on effect of the heating rate on the complete transformation temperature $T_{ct}$ of the Ti-6Al-1Mo-1V and BT6 alloys are presented in a semilog scale in Fig. 1. When the heating rate is increased from $1°C/s$ to $100°C/s$ increasing the upper temperature limit of $\alpha + \beta \rightarrow \beta$ transformation by $20°C$ to $100°C$ is observed. When heating at rates exceeding $100°C/s$ the displacement of transformation temperature limit is sharply slowed down which is attributed to replacing a diffusive mechanism of transformation by a nondiffusing one. As distinct from a high-purity and commercial grades of titanium characterized by $\alpha \rightarrow \beta$ transformation occurring via forming nuclei of $\beta$ phase and their further growth in two-phase alloys the $\alpha + \beta \rightarrow \beta$ phase trans-
formation results from a frontal boundary displacement of $\beta$ phases towards $\alpha$ phase (Fig. 2) during which atoms are randomly displaced on the distances lesser than interatomic ones or comparable with them. The displacement of interface on an isothermal and that with the rates lower than 100°C/s takes place when the necessary concentration conditions are achieved, the latter being created by diffusion of $\beta$ stabilizers in the region of phase interaction (i.e. a diffusive mechanism). With a nondiffusing mechanism $\alpha + \beta \rightarrow \beta$ transformation proceeds at actually complete suppressing the diffusive phase interaction. This transition can be considered a transformation in a one-phase system (1). In this case in the process of transformation there first occurs reconstruction of $\alpha \rightarrow \beta$ lattice during which an interface displaces in a frontal way and then the processes of diffusive levelling of $\beta$ stabilizer concentrations between an initial $\beta$ phase and that formed during $\alpha \rightarrow \beta$ transition proceed. The nondiffusing mechanism results in the formation of martensitic phases on quenching at an initial transformation stage. For example, in the Ti-6Al-1Mo-1V alloy the transformation at the heating rate of 100°C/s starts from 960°C and the martensitic phase is formed on quenching from 980°C (Fig. 2d). On quenching in a central part of $\beta$ interlayer $\beta$ phase having the concentration of $\alpha$ stabilizers which exceeds a critical one is formed and the martensitic transformation proceeds in $\beta$ phase regions which are adjacent to $\alpha$ lamellae and formed on heating during $\alpha + \beta \rightarrow \beta$ transition. Due to lack of time on high-speed heating for diffusive levelling of $\beta$ stabilizer concentration between an initial $\beta$ phase and that formed during $\alpha + \beta \rightarrow \beta$ transformation, in $\beta$ solid solution there appear regions with gradient distribution of alloying elements having concentration profiles controllable by varying the temperature and heating rate.

![Fig. 1: Temperature dependence of completing $\alpha \rightarrow \beta$ phase transformation on heating rate of Ti-6Al-1Mo-1V and BT6 alloys.](image-url)
Fig. 2: Microstructure of Ti-6Al-1Mo-1V alloy
(a) in initial state; and after heating with rates (b) 10°C/s to 960°C/s; (c) 10°C/s to 1040°C/s; (d) 100°C/s to 980°C/s.

The concentration distribution profiles of \( p \) stabilizing elements for a nondiffusing mechanism of transformation can be calculated by the second Fick equation.

The solution of this equation for the case of diffusion from a finite volume to a limited space can be presented in the form given in (9):

$$
c(x, \tau) = \frac{C_0 a}{t} + \sum_{n=1}^{\infty} \frac{2 C_0}{n \pi} \sin \left( \frac{n \pi a}{l} \right) \cos \left( \frac{n \pi x}{l} \right) \exp \left( - \frac{n^2 \pi^2 D \tau}{4l^2} \right)$$

Equ. 1

where \( c(x, \tau) \) is the concentration of an alloying element in a moment of time \( \tau \) at a point \( x \); \( C_0 \) is the initial concentration of an alloying element; \( l \) is the distance between particle centres of \( \alpha \) and \( p \) phases; \( a \) is a half \( \beta \) interlayer; \( n \) is a number of a series term; \( D \) is the effective coefficient of diffusion which can be calculated by the formula from (6):

$$
D = D_0 \exp \left( - \frac{Q}{RT} \right)
$$

Equ. 2

and where \( D_0 \) and \( Q \) are a preexponential factor and the activation energy, respectively. Using values of \( D_0 \), \( Q \) and \( C_0 \) given in (7, 8) and assuming \( l = 3 \mu m \) the concentration profiles of \( V \) distribution in the BTS alloys were calculated on computer on heating to 1025°C (\( \beta \) region) at various rates (Fig. 3). If the values of critical concentrations \( C'_{cr} \) and \( C''_{cr} \) separating the regions of metastable \( \alpha' \), \( \alpha'' \) and \( \beta \) phases were marked on given concentration profiles the volume fraction of the phases could be determined for any quenching temperature. Using a suggested mathematical model the specific structure states which are created in the
process of quenching from \( \beta \) region after high-speed continuous heating, e.g. dispersion of martensite and preserving a retained \( \beta \) phase, can be described to a sufficient accuracy.

Fig. 3: Concentration profiles of V distribution in BT6 alloy for 1 - initial state; 2 - \( R_h = 100^\circ C/s \); 3 - \( R_h = 300^\circ C/s \); 4 - \( R_h = 1000^\circ C/s \).

As it follows from the calculated data the length of martensitic acicular after heating at a rate of \( 100^\circ C \) up to \( 1025^\circ C \) (\( \beta \) region) and subsequent quenching should not exceed \( 5 \mu m \) due to the formation of regions with the concentration exceeding \( C_{cr} \) in \( \beta \) solid solution which is observed in a real alloy structure.

The formation of \( \beta \) stabilizer-riched and -depleted regions during heating results in formation of a set of \( \alpha' \), \( \alpha'' \) and \( \beta \) metastable phases on quenching. Under the conditions of high-speed heating up to the temperatures of \( \beta \) region an intensive grain growth is suppressed. The main parameters determining their growth are the temperature and heating rate. The relationships between \( \beta \) grain size, temperature and heating rate determined earlier by the authors can be presented in a semilog scale:

\[
\log D = \frac{1}{T} \quad \text{and} \quad \log D = \log R_h \quad \text{(Fig. 4)}.
\]

The existence of a linear dependence between the above parameters confirms the validity of the following relation:

\[
D = \frac{K_0}{R_h^n} \exp \left( -\frac{A}{T} \right) \quad \text{Equ. 3}
\]

where \( n \), \( K_0 \) and \( A \) are the coefficients depending on the chemical composition of the alloy. On strengthening treatment when using the high-speed heat treatment in the BT6, BT9 and BTJ-1 alloys having initial fine-grained structure the maximum strengthening occurs after quenchening from the tempera-
tures exceeding the limit of $\alpha + \beta \rightarrow \beta$ transformation (Fig. 5).

Fig. 4: Dependence of average grain diameter in BT9 on (a) temperature and (b) heating rate.

Fig. 5: Mechanical properties of (a) BT6 and (b) BT9 in quenched condition.

Due to the displacement of the phase transformation temperature the maximum strengthening shifts to the region of higher temperatures. With increasing the content of $\beta$ stabilizers in alloys of martensitic group after quenching subsequent to the high-speed electric heat treatment from optimal temperatures of $\beta$ region the tensile strength increases from 1250 MPa for the BT6 alloy and from 1400 MPa for the BT9 and plastic properties are maintained on a sufficiently high level ($\delta = 9$ to 12%, $\phi = 35%$ to 50%). The nature of favourable effect of the high speed heat treatment on structure and mechanical properties of titanium two-phase alloys is associa-
ted with the following factors: dispersion of martensitic constituents after quenching (increase of strength), suppressing grain growth and preserving a retained phase after quenching from the temperatures exceeding the upper temperature limit of transformation (maintaining a high plasticity).

Recrystallization in Two-Phase Titanium Alloys

The temperature ranges of recrystallization and phase transformation in two-phase titanium alloys are superimposed which complicates and disguises a real nature of changes in a dislocation structure.

In the process of hot deformation a cellular structure with 0.3 μm to 1 μm cell size is observed in α phase at the reduction of 30% and exceeding that (Fig. 6a). The 700°C isothermal annealing results in redistribution of dislocations and formation of some subgrains (Fig. 6b). In the range from 800°C to 900°C there occurs coarsening subgrains up to 1.5 μm to 2 μm by coalescence and migration of boundaries (Fig. 6c).

![Fig. 6: Microstructures of Ti-6Al-1Mo-1V after (a) deformation; subsequent furnace annealing (b) at 800°C, (c) 900°C and (d) high-speed heating (R_h=10°C/s) to 980°C.](image)

In 50% reduced alloys the temperatures of starting the primary recrystallization were 860°C, 870°C and 880°C for the BT3-1, BT9 and Ti-6Al-1Mo-1V alloys, respectively. By the temperature of completion of the primary recrystallization of alloys during the isothermal annealing one should imply the temperature of polymorphous α + β → β transformation because on annealing at temperatures 10 to 20°C below the upper boundary of transition the regions with lamellar α -
phase are retained where only polygonization proceeds. In two-phase alloys the necessary conditions for proceeding recrystallization in \( \alpha + \beta \) region are coagulation and spheroidization of phases.

Thus, a peculiar feature of "isothermal" recrystallization is existence of a wide temperature range from 120°C to 140°C which is associated with a retarding effect of \( \beta \) phase interlayers at the beginning and then as \( \alpha + \beta \rightarrow \beta \) transformation proceeds that of retained \( \alpha \) phase regions. On high-speed heating of two-phase alloys with an initial lamellar structure the process of phase coagulation is suppressed and the processes associated with the dislocation redistribution are subjected to a strong retarding action. Due to lack of time the processes of dislocation climbing fail to develop completely as they are restricted by a diffusive displacement of vacancies towards ascending dislocations. As a result the formation of subgrain boundaries pileups in cells proceeds at higher temperatures. The polygonization proceeds mainly in \( \alpha \) phase simultaneously with the development of \( \alpha + \beta \rightarrow \beta \) transformation. In those regions of \( \alpha \) phase where a polygonized structure has formed the transformation proceeds in such a way that \( \beta \) phase formed inherits a subgrain structure of \( \alpha \) phase (Fig. 6d). With high heating rates from 250°C/s to 300°C/s the primary recrystallization develops only in \( \beta \) region due to suppressing the dislocation distribution process. As a result of inheriting \( \alpha \) phase dislocation structure by \( \beta \) phase during \( \alpha + \beta \rightarrow \beta \) transformation the favourable conditions for creating new centres and subsequent development of recrystallization are created which results in refinement of a coarse-grained structure. After 30% reduction in \( \alpha + \beta \) region and subsequent high-speed heating to \( \beta \) region in the alloys tested a 12-to-16-fold refinement of initial coarse grain of a lamellar intergranular structure is achieved.

References
