TITANIUM ALLOYS' BEHAVIOUR UNDER VARIOUS LOADING CONDITIONS

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The paper deals with Ti-alloys on the base of $\alpha, \beta$ - solid solutions, two-phase $\alpha + \beta$ - alloys (Table I). The subject of investigation was the mechanism of deformation as well as the changing of mechanical properties under various loading conditions: static tension, impact bending, micro-impact affect (cavitation) in a wide temperature range from $-196^\circ C$ up to $+500^\circ C$.

Table I. Chemical Composition of Alloys Investigated

<table>
<thead>
<tr>
<th>Nos.</th>
<th>Alloy</th>
<th>w/o of alloying components</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>Al</td>
</tr>
<tr>
<td>1</td>
<td>Technical titanium</td>
<td></td>
</tr>
<tr>
<td>2</td>
<td>Ti5Al</td>
<td>2.7</td>
</tr>
<tr>
<td>3</td>
<td>Ti4Al</td>
<td>4.41</td>
</tr>
<tr>
<td>4</td>
<td>Ti6, 5Al</td>
<td>6.5</td>
</tr>
<tr>
<td>5</td>
<td>Ti5Al,2,5Sn</td>
<td>5.2</td>
</tr>
<tr>
<td>6</td>
<td>Ti4Al 3Mo 1V</td>
<td>4.23</td>
</tr>
<tr>
<td>7</td>
<td>Ti5Al 2Mo 2Cr</td>
<td>5.05</td>
</tr>
<tr>
<td>8</td>
<td>Ti2,5Al 7Mo</td>
<td>2.6</td>
</tr>
<tr>
<td>9</td>
<td>Ti8Mo</td>
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<tr>
<td>10</td>
<td>Ti3Al 8Mo</td>
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</tr>
<tr>
<td>11</td>
<td>Ti5Al 8Mo</td>
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</tr>
<tr>
<td>12</td>
<td>Ti2,5Al 11Mo</td>
<td>2.5</td>
</tr>
<tr>
<td>13</td>
<td>Ti9Mn</td>
<td>-</td>
</tr>
<tr>
<td>14</td>
<td>Ti3Al 6Mo 10Cr</td>
<td>3.18</td>
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The results of the research showed that plastic deformation under static tension at 20°C and in the field of negative temperatures takes place both by gliding and by twinning. Temperature lowering in the range of negative temperatures causes an increase of plastic deformation by twinning (1).

The possibility of realizing various mechanisms of plastic deformation determines the nature of mechanical properties change.

In technical titanium parallel with the growth of yield point and strength under temperature lowering in the range of negative temperatures relative elongation grows as well from 20% at 20°C up to 30% at -196°C. The growth of plasticity at temperature lowering results from, firstly, the possibility of essential development of twinning parallel with plastic deformation by gliding. Though twinning as such very little contributes to deformation, it provides crystal lattice re-orientation in separate volumes of metal and so creates conditions for consequent gliding, more favourably arranged planes for plastic deformation development by gliding (2). Secondly, the possibility of realizing various mechanisms of plastic deformation at lower temperatures results in increased strengthening during deformation due to bigger defects density of crystalline structure, which in its turn gives rise to increased resistance to big plastic deformations and destruction, and indirectly promotes the growth of relative elongation.

At temperatures from 20°C up to 500°C plastic deformation of technical titanium under static tension occurs principally by gliding, but faulting takes place as well. The latter reaches its greatest development at tension temperatures of 200-300°C.

The investigation of plastic deformation nature of α-alloys, Ti-Al and Ti-Al-Sn showed that aluminium and tin composing with titanium solid solution of substitution hamper the development of plastic deformation, particularly twinning and faulting processes. In Ti3Al, Ti4Al alloys, besides gliding, twinning is observed only in the negative temperatures range: at tension temperatures range of -120°C and -196°C. In higher alloyed α-alloys Ti6,5Al, Ti5Al 2,5Sn plastic deformation at static tension at room temperature as well as in the negative temperature range processes by sliding only. The fact that the development of plastic deformation in alloyed α-alloys is hampered causes the decrease of plastic properties in the negative temperature range. In alloyed α-alloys Ti6,5Al and Ti5Al 2,5Sn plastic deformation at tension temperatures of 20-500°C processes principally by gliding while in
The investigation of the influence of notches (holes, grooves) on the nature of plastic deformation under static tension of technical titanium flat specimens and \( \alpha \)-alloys showed that the change of stressed state under given loading conditions causes essential changes in the plastic deformation mechanism. With this plastic deformation begins at higher stress (the yield point is in the average 15 kg/mm\(^2\) higher in comparison with the specimens without stress concentrators and the elongation decreases approximately twice). Specimens with stress concentrators (with holes) of technical titanium an essential decrease of gliding in plastic deformation is observed, and at \(-196^\circ C\) deformation occurs principally by twinning. In \( \alpha \)-alloy Ti4Al under similar loading conditions the ability of plastic deformation development by gliding diminishes more sharply and already at \(-60^\circ C\) plastic deformation takes place principally by twinning. At lower temperatures the extent of plastic deformation developing by twinning decreases. The difficulty in development of plastic deformation by gliding and by twinning in specimens with notches in \( \alpha \)-alloys Ti-Al, Ti-Al-Sn, brings about a lowering of plastic properties and brittle destruction in the negative temperatures range (at \(-60^\circ C\)).

The study of plastic deformation mechanism of technical titanium and \( \alpha \)-alloys in more severe loading conditions (impact bending of notched specimens (ГОСТ 9454-60 type I) (3) in the temperature range of 20°C to \(-196^\circ C\) of technical titanium and \( \alpha \)-alloys processes principally by twinning. The degree of development of plastic deformation by twining with temperature lowering decreases. Impact viscosity of \( \alpha \)-alloys lowers from 8-10 kgm/cm\(^2\) at 20°C up to 3 kgm/cm\(^2\) at \(-196^\circ C\) (Fig. 1). At higher test temperatures 100-500°C in such loading conditions plastic deformation takes place principally by gliding, impact viscosity grows with temperature up to 16 kgm/cm\(^2\).

The study of plastic deformation nature in high-alloyed Ti3Al6Mo10Cr alloy, hardened at \( \beta \)-solid solution showed that with static tension of specimens without notches and with holes and render impact bending of notched specimens as well plastic deformation in \( \beta \)-phase in a wide temperature range from \(-196^\circ C\) up to 500°C processes by gliding. The increasing severity of loading (specimens with holes, impact bending of notched specimens, temperature lowering down to the range of negative ones) hampers plastic deformation by gliding causes reducing of plastic properties and brittle destruction in the
negative temperature range already at -70°C.

Tension diagrams from the specimens of Ti3Al6Mo10Cr alloy hardened at $\beta$-solid solution showed that at tension temperatures of 200, 300, 400°C diagrams have toothlike character which results from the Portven-Le-Chatelle effect (Fig. 2a). In this case the process of plastic deformation of instable $\beta$-solid solution in the temperature range of 200-400°C occurs an accumulation of atoms of admisive elements forming clouds of these atoms around the dislocations with consequent tearing dislocations out of the clouds under plastic deformation. With annealed specimens of the same heat the structure of which consists of $\beta$-solid solution separating $\alpha$-phase, such a toothlike nature of tension diagram was not observed (Fig. 2c). It is found that the dissociation of instable $\beta$-solid solution of Ti2.5Al7Mo, Ti3Al6Mo10Cr alloys in the process of tension at 550°C with separation of dispersed $\alpha$-phase causes growth of yield points and strength with simultaneous growth of elongation (Fig. 3). The growth of relative elongation results from diffusive plasticity displayed under transformation of $\beta\rightarrow\alpha$ in the tension process.

Plastic deformation of Ti3Al6Mo10Cr alloy in heat-treated state for maximum strength $G_\delta=150$ kg/mm$^2$, $\delta=4\%$ (hardening from 800°C + tempering at 450°C for 25 hours + holding at 560°C for 15 min) develops principally by faulting at tension temperatures of 20-300°C. In this case on tension curves jumps and derangements are observed (Fig.2b). At tension temperatures of 400°C and higher, when side by side with faulting plastic deformation by sliding gets its essential development jumps on the tension curves are not observed.

Availability of an instable $\beta$-phase in titanium alloys brings about substantial changes in the behaviour of alloys under various loading conditions (4). In the investigated alloys of Ti-Al-Mo at the temperature of the beginning of martensitic transformation lower than 300°C under hardening from $\beta$-field side by side with martensite, instable remanent $\beta$-phase is fixed able to dissociate during the process of deformation forming martensite and under heating in the range of 400-500°C with $\omega$-phase formation. It is stated that in these alloys non-stable remanent $\beta$-phase in the process of deformation under tension at 20°C and in the negative temperatures range desintegrates forming martensite which results in a low value of yield point ($\sigma_{y_2}=50-60$ kg/mm$^2$ at 20°C, 80-100 kg/mm$^2$ at -120°C), higher alloys strengthening during deformation, high ultimate strength (95-110 kg/mm$^2$ at 20°C, 115-130 kg/mm$^2$ at -120°C), keeping sufficient level of plasticity in the negative temperatures range (relative elongation
Fig. 1. The change of impact ductility of Ti-alloys depending on the testing temperature. 1) Ti5Al2.5Sn; 2) Ti2.5Al 7 Mo, annealed; 3) Ti2.5Al 7 Mo, hardening from 900°C; 4) Ti5Al 2 Mo 2 Cr, hardening from 860°C; 5) Ti3Al 2 Mo 2 Cr, hardening from 860°C, tempering 550°C.

Fig. 2. Tension diagrams of Ti3Al 6 Mo 10 Cr after various heat treatment. a) Hardening from 800°C; b) hardening from 800°C, tempering 450°C 25 hours holding at 560°C, 15 min; c) annealed at 650°C.
Fig. 3. Change of mechanical properties of Ti-alloys hardened from from β-field depending on testing temperature.
a) Ti2.5Al 7 Mo; b) Ti3Al 6 Mo 10 Cr.
10-14% at 20°C) (5). At lowering temperature from 20°C up to -196°C impact ductility decreases from 8 kg/cm² to 3 kg/cm². Such a change of properties results from the mechanism of plastic deformation. In this case plastic deformation mostly occurs on account of shearing process under transformation of β-phase into martensite along moving division boundaries between β-phase and martensite which results from low resistance to shearing at β-phase - martensite transformation. With the growth of the extent of deformation as the result of martensite transformation the defect density of crystalline structure increases. This provides an essential strengthening during the process of deformation and a high ultimate strength.

Under tension test of alloys with an unstable β-phase in the temperature range of 400-500°C a sharp fall of plastic characteristics takes place. A relative elongation under tension in temperature range of 20-300°C is maintained at the level of 12-14%, and under tension is temperature range of 400-500°C it falls up to 4-3%. The value of impact viscosity at lower test temperatures grow from 8 kgm/cm² at 20°C up to 14 kgm/cm² at 300°C and fall sharply up to 4 kgm/cm² in the test temperature range of 400-500°C (Fig.1). Such fall of plasticity and viscosity of alloys in the temperature range of 400-500°C results from desintegration of β-solid solution forming α-phase under the process of heating during 15 min. before the test.

The analysis of the metallographic picture of plastic deformation development and the destruction of Ti-alloys under microimpact loading in the process of cavitation affect (6) showed that the weakest resistance to cavitation affect is offered by α-solid solution. Under cavitation affect lines of gliding and twins appear. The destruction of α-solid solution has a viscous nature, develops from the grain boundaries, twins, and spreads over the whole grain (6). β-phase offers higher resistance to cavitation affect. Plastic deformation in β-phase under given loading conditions develops by gliding. At the initial period of cavitation an irregularity of plastic deformation in the volume of metal is observed: there are grains without any noticeable lines of gliding. The destruction of β-solid solution under micro-impact loading begins from the grains boundaries, lines of gliding, and

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6) The cavitation effect was accomplished on an impact erosion stand at the circumferential speed of the disc with the specimen fastened on it of 78 m/sec., the distance between the nozzle and the disc - 14 mm., and the water head - 0.28 atm.
finishes in shear of metal between the planes of gliding. The destruction of Ti28Ni, Ti25Co alloys with titanides separation begins with \( \alpha \)-solid solution. The destruction of alloys containing martensite parallel with \( \beta \)-solid solution processes from the martensite plates boundaries and spreads towards the latter. The destruction of alloys containing, besides \( \beta \)-solid solution, the \( \omega \)-phase takes place by shearing of metal between the planes of sliding and ends in brittle destruction.

It is stated that \( \alpha \)-alloys as their stableness against cavitation affect is concerned are at about the same level as steel IX18H9T (Fig. 5).

Alloys on the base of \( \alpha \)-solid solution with titanides separation in annealed state at hardness of 25-30 HRC as to their stability against cavitation are twice as stable as \( \alpha \)-alloys (7). In these alloys under hardening from \( \beta \)-field the \( \omega \)-phase is formed which results in increased hardness (50 HRC) and in brittle destruction during the process of cavitation. Stableness against cavitation of \( \alpha + \beta \) alloys phases is determined by relative correlation of structural components of \( \alpha \) - and \( \beta \) -phases, as well as by the degree of disperseability of \( \alpha \)-phase. The most stable against cavitation in annealed state is the high alloy Ti3Al6Mo10Cr, which can be explained by relatively high quantity of \( \beta \)-phase and by high dispersity of \( \alpha \)-phase. As a result of studying the titanium alloys behaviour under cavitation affect, with the instable solid solutions structure, it is stated that hardening alloys from two-phase field to the structure consisting of \( \alpha + \beta \) phases as compared with hardening from \( \beta \)-field to the martensite structure provides higher stability against cavitation.

Stability against cavitation of titanium alloys is greatly influenced by the ability of an instable \( \beta \)-solid solution to decompose under the micro-impact loading with martensite formation.

The formation of martensite during the process of cavitation was observed in Ti5Al8Mo, Ti3Al8Mo, Ti8Mo, Ti2,6Al7Mo alloys hardened from \( \beta \)-field to the structure of martensite with remanent instable \( \beta \)-phase. The stability against cavitation of such alloys is the higher the more of remanent \( \beta \)-phase is transformed into martensite during the process of microimpact affect. The latter is determined by the position of \( M_s \). For the investigated alloys \( M_s \) was in the temperature range from 370°C to 30°C. The lower the position of \( M_s \) in this range of temperatures, the more intensively the instable \( \beta \)-
phase is transformed into martensite in the process of cavitation and the higher is the stability against cavitation of the alloys. The destruction of the instable $\beta$ -phase with the formation of martensite at the initial stages of the process of deformation results in stress relaxation on the account of shear transformation and at the later stages of deformation provides an intensive strengthening due to additional increase of defect density as the result of shearing martensite transformation. This increase the resistance of the material to big plastic deformation and destruction.

There exists an analogy in the nature of strengthening and destruction of titanium alloys at tension and under cavitation effect. Technical titanium has a low yield point, shows a slight strengthening at room temperature both at tension and at micro-impact affect in the process of cavitation, and offers a small resistance to destruction at tension as well as under cavitation affect (Fig.5). Alloys containing a non-stable $\beta$ -phase (Ti5Al8Mo, Ti3Al8Mo, Ti18Mo, Ti2.5Al17Mo) dissociating during the process of deformation with martensite formation, possess a low yield point, but a marked ability to strengthen under plastic deformation (Fig.4a). The higher is the alloys strengthening coefficient, the higher resistance these alloys offer to destruction both at tension and at micro-impact affect. $\beta$ -phase alloys and two-phase $\alpha + \beta$ alloys, strengthened by heat treatment possess high values of yield points and strength with a sufficient plasticity which guarantees high resistance to destruction both at tension and under micro-impact affect, despite a slight strengthening during the process of deformation (Fig.5).

At a certain extent of alloyage with transient elements in titanium alloys hardened at $\beta$ -solid solution under micro-impact loading in the process of cavitation affect the dissociation of a nonstable $\beta$ -phase may take place forming $\omega$ -phase which results in brittle destruction of alloys during the test process. Alloys Ti9Mn, Ti3Mn1Fe1Cr1V pertain to such type of alloys. They have little losses under cavitation affect but undergo brittle destruction after 30 min of testing (Fig.5).

Applying strengthening heat treatment causing dissociation of $\beta$ -solid solution with formation of a disperse $\alpha$ -phase eliminates the chance of $\omega$ -phase formation under micro-impact loading during cavitation affect and guarantees growth of stability against cavitation about twice as big as in annealed condition. To highten stability against cavitation it is reasonable to harden two-phase alloys from the two-phase $\alpha + \beta$ field and to temper them at temperatures of 480-500°C. The optimum schedule of hardening and tempering are defined by
Fig. 4. Strengthening of Ti-alloys under static tension (a) and under cavitation affect (b).

α'-phase: 1) technical titanium; β-phase: 2) Ti9Mn; 3) Ti2,5Al 11 Mo; 4) Ti3Al 6 Mo 10 Cr; α'+β: 5) TiAl8Mo; 6) Ti3Al 8 Mo; 7) Ti8Mo; 8) Ti2,5Al7Mo.
FIG. 5. Stability of Ti-alloys against cavitation depending on phase composition:
1) $\alpha$-phase: technical titanium; 2) $\alpha + Ti_2$Me; 3) $\alpha + \beta$:
$Ti_5Al_2Mo2Cr, Ti_4A13Mo1V$; 4) $\beta + \alpha$ high-dispersed; 5) $\beta$-phase:
$Ti_2,5Al111 Mo, Ti_3A16Mo10Cr$; 6) $\beta + \omega$: $Ti_3Mn1Fe1Cr1V, Ti9Mn$.

the chemical composition of the alloy. Tempering at temperatures lower than the optimum ones does not eliminate the probability of $\omega$-phase forming, and results in brittle destruction. Tempering at temperatures higher than 500°C leads to destrengthening the solid solution, and to enlarging structural components ($\alpha$-phase) and lowers stability against cavitation.

**Conclusion**

The character of Ti-alloys behaviour render various loading conditions is defined by the stressed condition, by the possibility of plastic deformation processing under given loading conditions, by the phase composition and the dispersity degree of structural components, as well as by the possibility of dissociation of the instable $\beta$-phase with martensite formation of the $\omega$-phase and $\alpha$-phase during the process of loading.
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