STRENGTH AND PLASTICITY

Shock-Wave Properties and Deformation-Induced Structure of Commercial-Purity Titanium

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Abstract—The shock compressive wave profiles of commercial-purity titanium samples under different loading conditions have been measured. The spall strength of titanium as a function of the strain rate and temperature of deformation has been found. High-rate plastic deformation mechanisms have been studied. High-rate plastic deformation under the investigated loading conditions has been shown to occur by slip and twinning. The $\alpha \rightarrow \omega$ transformation has been established to begin at 12.2 GPa.

Keywords: titanium, shock-wave loading, dynamic strength, high-rate plastic deformation, structure

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INTRODUCTION

A rather large number of works are devoted to the investigation of titanium behavior under shock loading [1–12]. A major part of them deals with the investigation of the mechanisms of plastic deformation and fracture under plane shock-wave loading of commercial-purity VT1-0 titanium. Dynamic yield strength is less sensitive to material composition, structure, and test temperature than yield strength [1]. An increase in temperature from 20 to 600°C does not affect the resistance to high-rate plastic deformation [2]. An annealing-induced decrease in dislocation density reduces the hardness of the material, but increases its dynamic yield strength considerably [3]. High-rate deformation results in the formation of shear bands in the fracture region and adiabatic shear bands [4, 5].

Impact compression of titanium causes the $\alpha \rightarrow \omega$ transformation. According to different data [6–9], this transformation occurs at different pressures. The pressure depends on impurities in titanium, in particular, the presence of oxygen [8]. The thickness of the sample, the peak shock-wave pressure, the loading technique, and the temperature also affect the transformation pressure.

The majority of the data were obtained from the analysis of complete wave profiles. There are virtually no structural studies of VT1-0 titanium after different loading conditions. The structural data available in the literature are often limited only to X-ray diffraction patterns and optical images [5, 9].

Nevertheless, systematic structural studies can improve the importance of the available free-surface

velocity profiles. Therefore, this work aims to analyze the free-surface velocity profiles and to analyze the structure and microstructure of VT1-0 titanium samples after shock loading performed at various strain rates in the α -phase region.

EXPERIMENTAL

The object of the investigation was commercial-purity VT1-0 titanium. The total amount of additives in it (Si, Fe, O, H, N, C, other impurities) did not exceed 0.97 wt %.

The shock-wave experiments were carried out with the help of single-stage gas gun with a caliber of 44 mm, at the Russian Federal Nuclear Center—Zababakhin All-Russian Scientific Research Institute of Technical Physics. The collision velocity was measured by the electrocontact technique with an error of 0.2% [13].

A VT1-0 titanium rod 35 mm in diameter was the initial workpiece for samples and impactors. The samples were cut perpendicularly to the rod axis. The samples and impactors were cut in the shape of disks with a diameter of 34 mm and a thickness from 0.3 to 10 mm.

The velocity profiles of the free sample surface were analyzed to find the elastic-plastic properties of the material under study. The free surface velocity profiles were recorded at a time resolution of 2 ns and a measurement accuracy in the velocity range under study of not less than 1% using VISAR [14] and PDV [15] laser interferometric techniques.

The collision velocity was changed from 0.35 to 2.00 km/s. Tests were carried out in the pressure range

Fig. 1. Structure of VT1-0 titanium in the initial state.

3.9–26.5 GPa and the strain rate range in rarefaction waves 1×10^5 – 3×10^6 s⁻¹. The temperature of the samples was varied from –108 to 784°C. The samples were cooled in liquid nitrogen and heated with an induction heater.

The structure of the 4-mm-thick samples was examined. The structure of the samples was examined by X-ray diffraction (XRD) analysis, optical metallography, and transmission electron microscopy (TEM). X-ray diffraction analysis was carried out using a DRON-3 diffractometer with a diffracted-beam graphite monochromator in CuK α radiation. Optical metallography was conducted on a Neophot-2 metallographic microscope. Electron microscopic examination was performed using a JEM-200CX electron transmission microscope. Microhardness was measured using a PMT-3M tester at an indentation load of 0.98 N.

RESULTS AND DISCUSSION

According to the X-ray investigation of the phase composition of VT1-0 titanium, it consists of the α phase in its initial state. The diffraction peaks in the XRD patterns are quite narrow, indicating that there are no significant distortions in the structure.

The metallographic examination shows only grain boundaries in the sample (Fig. 1). There is no significant difference in the grain structure in the transverse and longitudinal sections of the initial workpiece. In both cases, the average grain size is 11–13 μ m.

The microhardness in the initial state is $H_\mu = 1837 \pm 55$ MPa.

Figure 2 shows the velocity-time profiles of the free surface of the VT1-0 titanium samples (the impact velocity is indicated next to each profile).

The wave profiles with an impact velocity less than 1.0 km/s show a splitting of the shock wave into an

Fig. 2. Velocity profiles of the free surface of the VT1-0 titanium sample were taken at room temperature and at different collision velocities (values are near profiles).

elastic precursor with compression stress behind its front equal to the dynamic elastic limit, and a plastic shock wave. A three-wave configuration of the shock wave exhibiting an $\alpha \rightarrow \omega$ transformation precursor was recorded in experiments at impact velocities of 1.34 and 1.45 km/s. The $\alpha \rightarrow \omega$ transformation in VT1-0 titanium begins at 12.2 GPa. This value agrees quite well with the data of [9]. The velocity of the plastic wave increases naturally with increasing shock compression pressure, whereas the time when the plastic wave reaches the target surface decreases regularly.

The resultant free surface velocity profiles were used to obtain the spall strength and shear stress relaxation in the material under study. The spall strength was calculated using the drop in the free surface velocity during the unloading of the shock-compressed state. The calculation was made taking into account the correction for distortion of the velocity profile due to the difference between the velocities of the spall pulse and the plastic wave in the unloading region [14].

Figure 3 shows the spall strength vs. strain rate of VT1-0 titanium. At low pressures, in the α -phase region, the spall strength increases monotonically until the strain rate reaches 7×10^5 s⁻¹ (indicated by circles and a solid line), at which the spall strength is ~ 4.3 GPa.

The experiments were divided into two groups in the region of high pressures. The first group included the experiments at nearly constant strain rates from $\sim 2 \times 10^5$ to $\sim 3 \times 10^5$ s⁻¹. In this group, different completeness of $\alpha \rightarrow \omega$ transformation can be achieved by varying the impact velocity, and the spall strength increases noticeably from ~ 3.7 to ~ 4.9 GPa ($\alpha \rightarrow \omega$ group is indicated by triangles and dotted lines). The second group includes the experiments performed at an impact velocity of ~ 2.0 km/s. In this case, $\alpha \rightarrow \omega$ transformation was complete. The spall strength grew

Fig. 3. (a) Spall strength of VT1-0 titanium as a function of the deformation rate and $\alpha \rightarrow \omega$ transformation degree. (b) The maximum deformation rate as a function of the pressure amplitude. Room-temperature loading.

from ~ 4.9 to ~ 6 GPa when the strain rate was increased from $\sim 3 \times 10^5$ to $\sim 3 \times 10^6 s^{-1}$ (ω -phase group).

Analysis of the temperature effect showed that the spall strength decreases with increasing temperature. The spall strength at the initial sample temperature of $784^\circ C$ and strain rate of $\sim 2 \times 10^5 s^{-1}$ is 2.4 GPa, which is 40% lower than its value under normal conditions. When the temperature decreases to $-108^\circ C$, the spall strength remains almost the same. In general, these data agree with the results of [10].

The shock-wave loading at impact velocities below 1.0 km/s results in a two-wave configuration of the shock wave, the material remains to be an α -phase one, and the strain rate increases during shock compression (Fig. 3b). Therefore, the structural studies were carried out on the samples retained after room-temperature loading at impact velocities $V = 0.35$ and 0.57 m/s; the strain rate increased from $\sim 3 \times 10^5$ to $\sim 1 \times 10^6 s^{-1}$.

The loading of VT1-0 titanium at a velocity of 0.35 km/s (peak pressure of 3.9 GPa) forms a spall crack in the cross-section approximately at the middle of the sample thickness. An increase in the pressure to 6.4 GPa at the cost of increasing the impact loading velocity to 0.57 km/s increases the cross-sectional size of the main crack significantly. It passes through the middle part of the sample in the same way as it did in the previous case, but the crack is more developed.

Electron microscopic study shows that the pulse impact at $V = 0.35$ km/s results in high-rate plastic deformation of VT1-0 titanium through slip and twinning (Fig. 4). The dislocation structure of the sample is nonuniform. In addition to regions containing a large number of uniformly distributed dislocations (Fig. 4a), there are regions with specific pileups of elongated dislocations. For example, two pileups can be seen in Fig. 4b.

They spread almost parallel to each other. Moreover, some dislocations located between them are also

elongated in this direction. These pileups are in fact cross sections of extended planar dislocation pileups.

There are only several regions with microtwins (Figs. 4c, 4d). Typically, almost all these twins contain dislocations inside themselves. Twinning in α titanium during plastic deformation can occur along six planes [16]. The predominant type of twinning depends on both the loading scheme and the temperature of deformation. The twinning planes at room temperature are the $\{10\bar{1}2\}$, $\{11\bar{2}1\}$, and $\{11\bar{2}\bar{2}\}$ planes. The twinning plane in electron microscopic study is determined by the known direction of its trace on the image and the known axes of the matrix and twin zones. Figure 4d shows the matrix with the $[211]$ zone axis. The absence of all twin reflections makes it impossible to determine its zone axis. Therefore, the twinning plane was found only using the matrix orientation. To do this, we identified the direction of the intersection of all possible twinning planes with the foil plane. A comparison of these directions with the twinning plane trace direction in Fig. 4d shows that only the trace from the $(\bar{1}102)$ plane coincides with it.

An increase in the impact velocity to 0.57 km/s causes a significant increase in the dislocation density as compared to the dislocation density after the low-velocity loading (Fig. 5). The dislocation number density was estimated to be $\sim 10^{16} m^{-2}$. The dislocations are distributed mainly homogeneously, but there are regions with a low dislocation density, where dislocations are elongated along crystallographic directions (Figs. 5a, 5b). Similar to the structure formed during the low-velocity loading, the titanium structure in this case exhibits bands with a large number of dislocations (Fig. 5b).

A subgrain structure (Figs. 5c, 5d) forms rapidly under these loading conditions and an increase in the temperature results in partial polygonization.

The number density of microtwins increases slightly in comparison with that under low-velocity

Fig. 4. Microstructure of VT1-0 titanium after pulsed room-temperature impact at $V = 0.35$ km/s. The twinning plane is indicated by the dashed line: (a) dislocations, (b) dislocation pileups, and (c), (d) microtwins.

loading (Figs. 5e, 5f). In many regions, microtwins stop growing at the boundaries of the polygonal structure that is being formed.

The microhardness was measured along the cross section of the samples. The microhardness after pulse loading at $V = 0.35$ km/s increases to 2016 ± 80 MPa. The microhardness increment is 9.7%. The high-velocity loading at $V = 0.57$ km/s increases the microhardness of the alloy to an average value of 2098 ± 69 MPa (by 14.2% relative to the initial microhardness). After both loading regimes, the microhardness level within the measurement error does not change along the thickness of the disks.

The wave profiles for VT1-0 titanium mainly agree with the data from [1, 3, 12] obtained under similar loading conditions. In addition, we found that the strain rate-dependent spall strength changes differently at different pressures; namely, it grows monotonically with the strain rate at low pressures, whereas it grows faster at high pressures. This behavior of the spall strength can be attributed to the formation of the ω phase in this pressure range.

The appearance of the spall cracks indicates that they formed through the nucleation of a large number of pores, their growth, and subsequent pore coalescence.

In contrast to the work [5], the formation of adiabatic shear bands is not observed under the loading

conditions we used. Their formation was found [17] to occur at higher shock wave amplitudes.

High-rate plastic deformation occurs by slip and twinning under all loading conditions under the investigation. There are few twins on the microlevel, and their number increases insignificantly with increasing loading velocity. Therefore, slip makes the main contribution to stress relaxation. The dislocation density increases greatly with increasing loading velocity, reaching values exceeding 10^{16} m^{-2} at impact velocity $V = 0.57$ km/s. The dislocation density was noted [18] to be one of the main factors responsible for the formation of a cellular dislocation structure. However, no evidence of the cellular structure formation is detected, despite high dislocation density in the VT1-0 titanium samples under study. This is probably because no fixed Lomer–Cottrell locks are formed in the titanium structure during high-rate plastic deformation. No cellular dislocation structure has been shown [19] to form in the copper single crystal after loading by spherically converging shock waves if there are no fixed Lomer–Cottrell locks in the structure. In titanium, slip mainly proceeds along prismatic and pyramidal planes. The absence of a cellular structure in the investigated samples suggests that slip along these planes does not cause the formation of fixed Lomer–Cottrell locks.

Fig. 5. Microstructure of VT1-0 titanium after pulsed room-temperature impact at $V = 0.57$ km/s: (a), (b) dislocations elongated along crystallographic directions, (c), (d) subgrain structure, and (e), (f) microtwins.

In addition to uniformly distributed dislocations, there are also plane dislocation pileups in the deformed titanium structure. They form due to densely localized deformation resulted from plastic deformation occurring in several closely located slip planes. The dislocation density in these pileups is very high. Edge dislocations and edge components of mixed dislocations reach grain boundaries when stresses are high, therefore, mainly screw dislocations remain in the localized deformation band, since the velocity of edge dislocations many times exceeds the velocity of screw dislocations.

The spall strength increases with increasing strain rate, due to both a decrease in the tensile stress zone width and an increase in the dislocation density. The spall strength decreases with increasing temperature,

because plastic deformation starts at low effective stresses.

CONCLUSIONS

The effect of shock compression on the rheological properties and deformation structure of VT1-0 titanium was studied.

We found that the $\alpha \rightarrow \omega$ transformation began at 12.2 GPa.

The spall strength of titanium depends on the strain rate and changes differently; namely, at low pressures, it grows monotonically with the strain rate, and at high pressures it increases substantially in the ω -phase formation region, depending on the $\alpha \rightarrow \omega$ -transformation completeness.

The recorded dynamic elastic limit and the spall strength decrease at a temperature of 784°C.

Spall cracks form through the nucleation, growth, and subsequent coalescence of pores.

High-rate plastic deformation under the investigated loading conditions was shown to occur by slip and twinning. Twinning occurs over the $\{10\bar{1}2\}$ plane. No adiabatic shear bands were been observed.

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