

TITANIUM AND
TITANIUM ALLOYS
Scientific and Technological Aspects
Volume 3

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 Springer

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Volume 3

The Metallurgical Society of AIME Proceedings
published by Plenum Press

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TITANIUM AND TITANIUM ALLOYS

Scientific and Technological Aspects

Volume 3

Library of Congress Cataloging in Publication Data

International Conference on Titanium, 3d, Moscow State University, 1976.
Titanium and titanium alloys.

Organized by the Academy of Sciences of the USSR, in association with the Metallurgical Society of AIME and others, and held at Moscow State University, May 18–21, 1976.

Includes index.

1. Titanium—Congresses. 2. Titanium alloys—Congresses. I. Williams, James Case, 1938- II. Belov, Aleksandr Fedorovich. III. Akademiia nauk SSSR. IV. Title.

TN693.T5157 1976 669'.732 79-9156

ISBN 978-1-4757-1760-0 ISBN 978-1-4757-1758-7 (eBook)

DOI 10.1007/978-1-4757-1758-7

Proceedings of the Third International Conference on Titanium, organized by the Academy of Sciences of the USSR, in association with The Metallurgical Society of AIME, American Society for Metals, the Metals Society (London), and the Japan Institute of Metals, in Moscow, on May 18–21, 1976.

© 1982 Springer Science+Business Media New York

Originally published by Plenum Press, New York in 1982

Softcover reprint of the hardcover 1st edition 1982

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DECOMPOSITION OF THE MARTENSITE IN TWO-PHASE TITANIUM ALLOYS

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The present work is devoted to an electron microscope study of the decomposition of hexagonal martensite in $\alpha+\beta$ titanium alloys. As previous investigations have shown, martensite decomposition in titanium alloys is heterogeneous in character.^(1,2) In this connection it is of interest to study the martensite substructure and its effect on the decomposition kinetics, dispersity and distribution density of precipitating phases.

For this investigation, the alloy VT9 (Ti-6.5Al-3.5Mo2Zr-0.2Si) quenched from the β -field was used. Upon quenching from 1050°C, hexagonal martensite forms and a very small quantity of β -phase is retained. Observations show the internal structure of martensite plates to be varied. One can note the following kinds of internal structure of martensite:

1. Rows of parallel dislocations (Fig. 1);
2. Blocks of dislocation origin with a slight disalignment (Fig. 2);
3. Complicated dislocation pile-ups and networks (Fig. 3);
4. Twins with a twinning plane $\{1011\}_{\alpha}'$ (Fig. 4).

According to the existing phenomenological theory of crystallography of the martensitic transformation, martensite forms by an invariant plane deformation which generally consists of a homogeneous lattice-invariant shear.⁽³⁾

Principal distortions for the transformation of b.c.c. to h.c.p. lattice are:

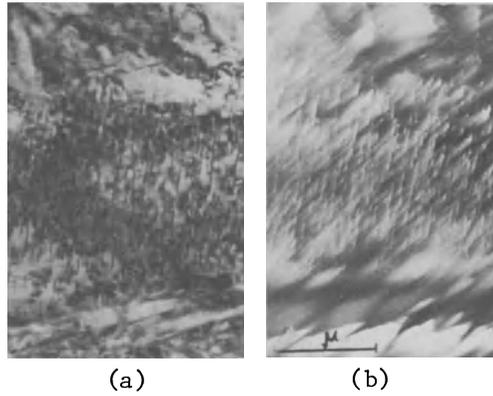


Fig. 1: a) Rows of dislocations in martensite. Quenched from 1100°C (1hr) + tempered at 550°C for 1hr. Z.N. $[1322]_{\alpha}$; b) Precipitates of β -phase on the rows. D.F. micrograph from $(020)_{\beta}$ reflection. $\times 29000$

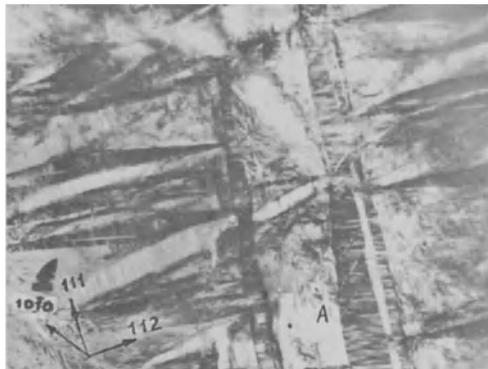


Fig. 2: Block structure of martensite. Quenching from 1050°C 15 min. Z.N. $[110]_{\beta} || [0001]_{\alpha}$. A-twins.

$$\varepsilon_i = \eta_i - 1,$$

where

$$\eta_1 = \frac{a_\alpha}{a_\beta} \sqrt{3/2}, \quad \eta_2 = \frac{a_\alpha}{a_\beta}, \quad \eta_3 = \frac{c_\alpha}{a_\beta} \sqrt{2}$$

and where a_α and c_α = lattice parameters of α -phase (h.c.p.),

a_β = lattice parameter of β -phase (b.c.c.).

For the investigated alloy $a_\beta = 3.299$, $a_\alpha = 2.949$ and $c_\alpha = 4.677\text{\AA}$ and $c/a = 1.586^*$; $\varepsilon_1 = 9.5$; $\varepsilon_2 = 10.5$ and $\varepsilon_3 = 0.3\%$.

Since $\varepsilon_3 = 0$ is small, it can be taken as 0. At $\varepsilon_3 = 0$ no additional lattice-invariant shear is required as the lattice deformation itself proves to be an invariant plane strain. In this case it can be accomplished by a simple shear with an invariant (habit) plane.

Using the computation of habit plane indices employed in (3) for the case when $\varepsilon_3 = 0$ it was found that for the investigated alloy they should be $\{575\}_\beta$. Loci of habit plane normals were determined by trace analysis. They are plotted on the Figure 5 assuming the Burgers orientation relationship is fulfilled. Experimental results are in satisfactory agreement with the computations. Therefore, one can actually assume that during $\beta \rightarrow \alpha'$ -transformation the only deformation in the investigated alloy is the lattice deformation. It appears to be close to the deformation for the lattice change proposed by Burgers for $\beta \rightarrow \alpha$ -transformation in Zr. It is a shear in $\langle 11\bar{1} \rangle$ direction on a $\{112\}$ plane.

Proceeding from the fact that the deformation of $\beta \rightarrow \alpha$ -transformation is identical to the shear for the $\beta \rightarrow \alpha$ lattice change according to Burgers, one can suppose that accommodation of stresses between α' -crystals and the β -matrix occurs by formation of translation blocks with the output of dislocations on their boundaries. A similar mechanism of translation block formation is offered for the case of α -phase growth.⁽⁴⁾ Rows of parallel dislocations observed in martensite plates of the investigated alloy (Fig. 1) appear to be connected with the formation of such blocks in the $\beta \rightarrow \alpha'$ -transformation. The following testifies in favor of this suggestion:

1. Trace analysis showed the dislocations to be located on planes $\{10\bar{1}0\}_\alpha \parallel \{112\}_\beta$ along $\langle 1\bar{2}10 \rangle_\alpha \parallel \langle 11\bar{1} \rangle_\beta$, so that they are likely to be screw dislocations with $\beta = 1/3 \langle 1\bar{2}10 \rangle$;

*Parameter values were determined by X-ray method.



Fig. 3: Dislocation pileups and twins in martensite. Quenched from 1050°C (1hr) + tempered 600°C , 2hr. A - Dislocation pileups, B - twins. $\times 17000$

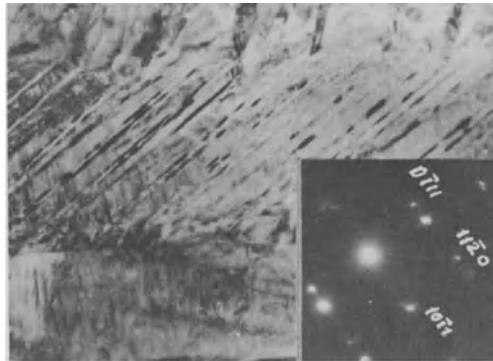


Fig. 4: β -phase precipitates on twin boundaries in martensite. Quenched from 1050°C (1hr) + tempered at 600°C for 2hr. Z.N. $[001]_{\beta} \parallel [110]_{\alpha}$, Z.N. of twins $[110]_{\alpha}$. $\times 37000$

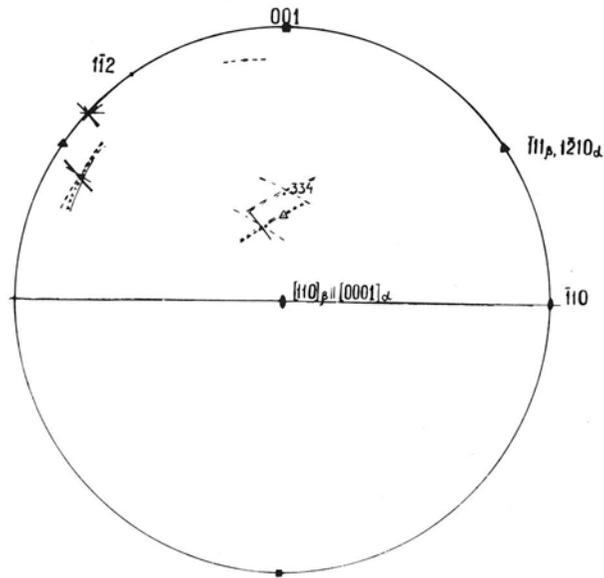


Fig. 5: Trace analysis of habit planes of martensite plates. --- and === quenching after β -deformation.

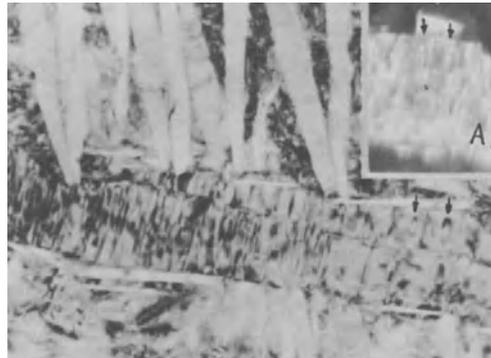


Fig. 6: Dislocation loops in the alloy after quenching from 1050°C for 1hr + tempering at 550°C for 100hr. A - D.f. micrograph from $(10\bar{1}0)_{\alpha}$ reflection. x50000

2. The amount of shear necessary for $\beta \rightarrow \alpha'$ -transformation computed according to Wechsler-Otte⁽⁵⁾ is about 0.026 of the shear for slip on $\{1010\}$ along $\langle 1210 \rangle$. This amount of shear corresponds to the formation of a single dislocation every 80 planes. Spacing between the dislocations then must be roughly 200\AA , which agrees with that observed (Fig. 1).

The formation of translation blocks can occur in spite of the fact that it does not completely remove the stresses at the martensite plate-matrix boundary plane. For the complete relaxation of stresses to occur, a little shear along this plane and tension in the direction normal to it are also necessary. As these deformations are small (below 3%), they can be accommodated at the expense of elastic deformation of the plate and the matrix.

In general, during martensitic transformation a complicated stressed state of the material is formed due to the collision of martensite plates during their formation, anisotropy of linear expansion of α -phase and the development of stresses at the martensite-matrix boundary surface. Therefore one can understand the observed breakdown of martensite crystals into disoriented blocks (Fig. 2) or the formation of dislocation pile-ups (Fig. 3,A) and twins in them (Fig. 3,B). All of this leads to relaxation of stresses. The process of breakdown into disoriented blocks is similar to the formation of disoriented microbands in the presence of an inhomogeneous stressed state in deformation of mono- and polycrystals.⁽⁶⁾

On the basis of the Bowles-Mackenzie⁽⁷⁾ theory, the lattice-invariant shear for transformation from b.c.c. to an orthorhombic (or h.c.p.) lattice must occur by twinning on $\{110\}_{\text{b.c.c.}}$ in an irrational direction (class of solutions A). The shear value on the plane $(110)_{\beta} \parallel (1\bar{1}01)_{\alpha'}$ estimated according to Wechsler-Otte⁽⁵⁾ amounts to ~ 0.03 of a twinning shear. Therefore, in this alloy, twins must not occur. However, a good many internal twinned martensite plates are observed fairly often. Sometimes both twins and blocks of dislocation origin are observed in large plates simultaneously (Figs. 2 and 4). Presumably in this case twinning is a consequence of deformation due to the shock of martensite crystals formed later. That twins in martensite of the investigated alloy do not result from the transformation, but from the deformation is confirmed by the following:

1. Contrast of blocks continues behind the twins and is sometimes visible in the twins proper (Figs. 2,A and 4).
2. Deformation of α -titanium at room temperature does not occur by twinning on $\{10\bar{1}1\}$, but at elevated temperatures, as shown in (8), this very type of twinning deformation is possible. M_s of



Fig. 7: β -particles: A-on dislocation networks, B - on martensite plate boundaries. Quenched from 1050°C (1hr) + tempered at 600°C for 2hr.



Fig. 8: Ordering in α -plates. Quenched from 1050°C (1hr) + tempered 500°C , 100hr. Z.N. $[10\bar{1}]_{\beta} || [\bar{5}14\bar{3}]_{\alpha, \alpha_2}$. x30000

the investigated alloy is much higher than the ambient temperature.

Heating the quenched alloy initiates the decomposition of martensite. At low temperatures and short durations of tempering, microstructural changes were not observed. After tempering at 550°C for 5 hr, X-ray and electron microscope study revealed the beginning of β -phase precipitation from α' -martensite. β -phase particles seem to be more distinct and are seen earlier at twin boundaries (Fig. 4). The twinning plane $\{10\bar{1}1\}$, that is the habit plane of the twins, is parallel to $\{110\}_\beta$, where the α - and β -phase lattices are in good matching. It is a favorable site for the germination of β -phase. β -precipitates appear also at the boundaries of smaller martensite plates and on rows of dislocations with $\vec{\beta} = 1/3 \langle 1\bar{2}10 \rangle$ on $\{10\bar{1}0\}_\alpha \parallel \{112\}_\beta$ planes in larger plates (Fig. 1). After tempering at 550°C for 100 hr the formation of dislocation loops occurs in some α -plates in which disoriented blocks were observed after quenching (Fig. 6). Elevation of temperature and increased duration of tempering results in the formation of β -phase in the loops. β -phase is also precipitated at dislocation pile-ups in large α -plates and at martensite plate boundaries (Fig. 7).



Fig. 9: Recrystallized α -phase. Quenched from 1050°C (30min) + tempered 600°C, 100hr. Z.N. $[\bar{1}0\bar{1}]_\beta \parallel [5\bar{1}4\bar{3}]$. x30000

At 550°C for 100 hr, conditions for the development of an ordering process are created as a result of martensite depletion in Mo and the enrichment in Al during the precipitation of β -phase. Faint superstructural reflections are seen on the electron diffraction patterns taken from large plates (Fig. 8). Ordering is of Ti_3Al type. Superstructural reflections remain on the electron diffraction patterns after 50 hr at 600°C and 5 hr at 650°C.

Mechanical tests of the alloy showed that the strength and the relative elongation are lowered at that stage of tempering (600°C, 5 hr) when β -phase precipitates are seen in all martensite plates on dislocations, twin boundaries and blocks; β -phase also precipitates on plate boundaries and an ordered α_2 -phase is formed in the plates. With a further increased duration of tempering and the elevation of temperature β -phase particles begin to coagulate, forming β -islands at the boundaries of blocks and plates proper. Simultaneous recovery and recrystallization processes take place in the martensite plates. The plate body becomes free from dislocations, becoming a monolith. At the tempering temperature 650°C, recrystallization begins already after 30 min. The process of recrystallization in the former martensite plates occurs mainly by means of coalescence. Therefore, inside the recrystallized α -plates, β -phase islands are left. The most obvious is the gradual disappearance of boundaries between parallel α -plates with the same orientation (Fig. 9) which merge together in a single body with β -phase veins. The recrystallization results in a further decrease of strength and elevation of tensile ductility.

Basic elements of martensite substructure of the investigated alloy are blocks and twins. The β -phase precipitation as a continuous film on twin boundaries and on the martensite plate boundaries being about deterioration in mechanical properties. The precipitation at the dislocation pile-ups and networks remains highly dispersed for a long time. By changing the martensite substructure one can probably prevent a sharp decrease of tensile ductility of the alloy in the state strengthened by heat treatment. In reference 9 it is shown that the deformation of martensite improves the mechanical properties of alloy Ti-6Al-4V after tempering, especially at 600°C. Evidently the introduction of additional dislocations as well as the formation of dislocation networks and pile-ups in the martensite results in optimum tensile ductility of the alloy in the state strengthened by heat treatment.

Deformation in the β -field favors the formation of complicated dislocation networks and pile-ups in the martensite and also makes it fine as a result of the inheritance of parent β -phase dislocations.⁽¹⁰⁾ Hence β -deformation will lead to better tensile ductility of the alloy in the state strengthened by heat treatment in comparison with the properties of the undeformed alloy. A necessary condition for the favorable influence of β -deformation is considerable deformation without recrystallization.

REFERENCES

1. Poljak, E. V. and Sokolova, A. Yu., Structural transformations in two-phase titanium alloys, Sb. "Novij konstrukcionnij material titan," "Nauka," M., 1972, p. 66.
2. Williams, J. C. Critical Review. Kinetics and phase transformations, "Titanium Science and Technology," Plenum, New York, 1973, 3, 1433.
3. Kelly, A. and Groves, G. W., Crystallography and crystal defects. Longman, London, 1970.
4. Murzayeva, G. V. and Lerinman, R. M., Electron microscopic investigation of metastable β -phase decomposition in titanium alloy TS6. "Fiz. metal. metalloved.," 1970, 28, 813.
5. Wechsler, M. S. and Otte, H. M., The generalized theory of the martensitic cubic to orthorhombic phase transformation. "Acta Met.," 1961, 9, 117.
6. Urusovskaja, A. A., Formation of regions with disoriented lattice at deformation of mono- and polycrystals. "Sb. Nekotorie voprosi fiziki plastichnosti kristallov," Izd. AN SSSR, 1960, 75.
7. Mackenzie, J. K. and Bowles, J. S., The crystallography of martensite transformations-IV. Body centered cubic to orthorhombic transformations. "Acta Met.," 1957, 5, 137.
8. Nath, B. and Lorimer, G. W., The effect of thermomechanical history of the structure and properties of a Ti-6Al-4V alloy. Proceedings of the Third International Conference on the Strength of Metals and Alloys., Cambridge, England, 1973, 1, 123.
10. Trenoghina, T. L., Yelkina, O. A., Lerinman, R. M., Zvereva, Z. V., Beloborodova, A. T. and Makhnev, E. S., Electron microscope study of effect of thermomechanical conditions of deformation on the structure two-phase $\alpha+\beta$ -titanium alloys. "Fiz. metal. metalloved.," 1975, 40, 1227.

THE INVESTIGATION OF THE TITANIUM STRUCTURE
AFTER SHOCK WAVE LOADING

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INTRODUCTION

The analysis of the previous data on phase transformations under shock wave loading⁽¹⁾ shows that the same polymorphic transitions occur in short duration conditions of dynamic compression as in static high-pressure experiments. Until quite recently the $\alpha - \omega$ transformation in titanium was not detected during shock loading.

Under static pressures titanium undergoes $\alpha - \omega$ transformation between 40 and 90kbar.^(2,3) But no phase transition was noticed in that range at shock measurements. The kink on the D - U diagram occurred at 175kbar⁽⁴⁾ and was correlated with the $\alpha - \beta$ transition. This was verified by observing β -phase in the titanium samples after unloading. Retention of β -phase also detected by Pashkov and Polyakova, as described in reference 5. The structure and property changes in Ti after explosive shock loading were also correlated with the $\alpha - \beta$ transformation.⁽⁶⁾ The occurrence of the Ti $\alpha - \omega$ transition during dynamic compression was established for the first time in⁽⁷⁾. It was also shown that considerable quantity of ω -phase may be retained after unloading.

Because of the presence of ω -phase at atmospheric pressure it became possible to study the structure and phase composition of titanium as a function of loading conditions. The aim of the present work is to gain additional knowledge of the mechanism and characteristics of the $\alpha - \omega$ transformation at shock wave loading.

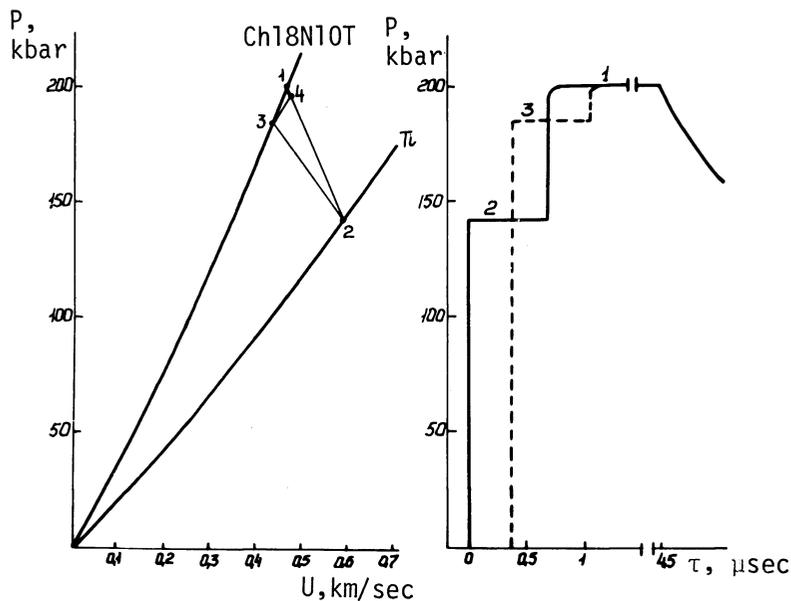


Fig. 1: Shock loading scheme for the titanium sample in the container of stainless steel Ch18N10T. a) Pressures (P) and particle velocities (U) in the container and the sample, 2,4,1- P-U states on the entrance interface (steel-titanium); 3,1 - P-U states on the back interface (titanium-steel). b) Pressure (P) vs time (T) for the entrance (—) and back (---) sample surfaces; pressures in state 4 and 1 are practically the same.

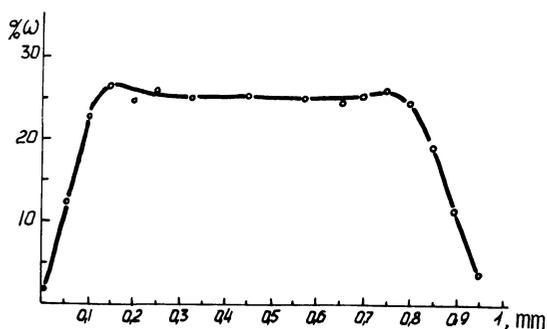


Fig. 2: β -phase distribution along cross section of titanium sample after loading to 200 kbar at the initial temperature 120°K (steel container).

EXPERIMENTAL PROCEDURE

Titanium VT-1-0 in the form of a cylinder (16mm diameter by 2mm thick) and iodide grade titanium (16mm diameter by 1mm thick) were used for this investigation. The samples were enclosed in a cylindrical container of stainless steel CH18N10T. Plane shock waves of 120, 200, 350 and 500kbar were produced by the method of contact exposure. The waves were produced by the impact of the metal driver plate accelerated by an explosive charge with the container top surface. This procedure was used earlier⁽⁷⁾ and described in detail.⁽⁸⁾

Since the sample and container materials are of different dynamic impedance, loading to the nominal pressure of 120, 200, 350 and 500kbar was done by stages. P-U and P-T loading diagrams (200kbar) for titanium are given as examples in Fig. 1. It is seen from Fig. 1 that the pressure close to maximum initial pressure in the container (state 1) is achieved in the sample in $\sim 1\mu\text{sec}$ while the whole compression impulse duration is $\sim 4.5\mu\text{sec}$. Similar behavior is observed at the other pressures. In several experiments a titanium container was used to get singlewave loading. In that case only state 2 (Fig. 1) with $\sim 4.5\mu\text{sec}$ duration is realized and pressures of 90, 140, 265 and 390kbar passing through the sample are attained in the first shock wave.

X-ray methods were used to investigate ω -phase content by measuring $(011)_\alpha$ and $(110)_\omega$ line intensities. Surface layers of 0.1-0.15mm thick were removed by etching in hydrofluoric acid. For transmission electron microscopy, discs of 4mm diameter were cut from the samples, mechanically thinned to 0, 1-0, 2mm and then chemically polished in the 1:4 mixture of hydrofluoric and nitric acids until suitable thin foils were obtained.

EXPERIMENTAL RESULTS

No ω -phase was detected in titanium samples after shock wave loading of 120-500kbar (steel container) at the initial temperature of 290°K. At the lower initial temperature of 120°K ω -phase was 6, 25, 39 and 54% at 120, 200, 300 and 500kbar, respectively.

After singlewave loading of 90kbar (titanium container, initial temperature 120°K) traces of ω -phase were detected; 140kbar loading resulted in 25% ω -phase. In the experiments with singlewave loading to 265 and 386kbar three titanium samples (each 2mm thick) were enclosed in container, 0,1mm titanium foil being put between them. No ω -phase was detected in these samples. During these experiments containers were subjected to considerable heating, which is evidenced by a blue discoloration on the surface.

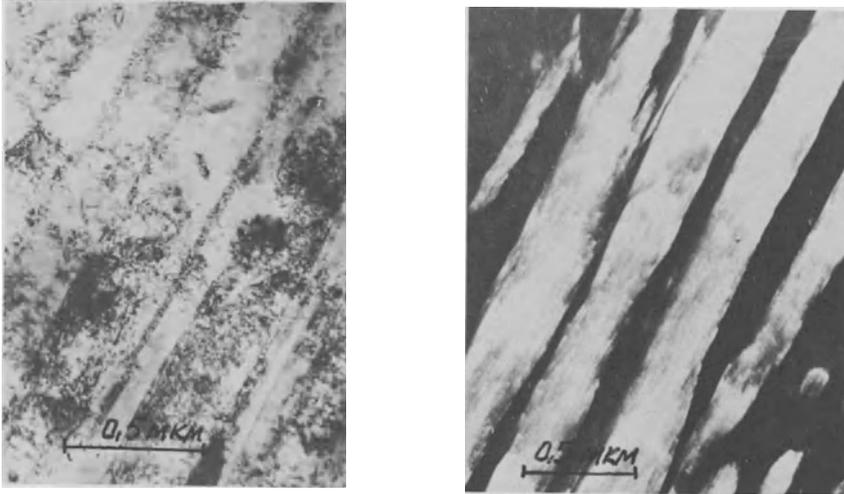


Fig. 3: Structure of shock-loaded titanium (140 kbar, 120^o K, titanium-container)
 a) α - and ω -crystals.
 b) Dark-field image in ω -phase pattern spot.

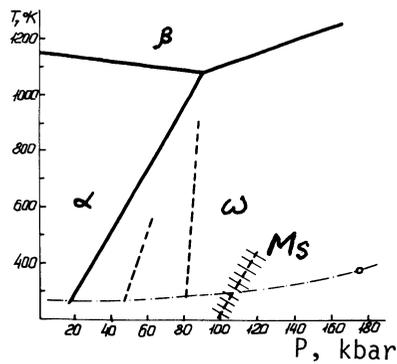


Fig. 4: T-P diagram of Ti.
 ——— phase equilibrium lines (Ref. 12);
 -.-.- Hugoniot for Ti, o-phase transition point (Ref. 4);
 --- isothermal $\alpha \rightarrow \omega$ transformation lines (Ref. 2,3);
 ##### line of the athermal martensite $\alpha \rightarrow \omega$ transformation (M_s).

Four experiments at 350kbar and 120°K were done to determine phase composition stability. ω -phase content was 34, 20, 21 and 24%. The second loading 350kbar, 120°K of the samples with 21 and 24% ω -phase resulted in ω -phase increase to 34 and 35%, respectively.

ω -phase content along the cross-section of the titanium sample was examined by means of X-ray studies. Several layers of 0,05mm were consecutively removed and the ω -phase distribution after 200kbar loading at 120°K is shown in Fig. 2. Similar behavior was observed in other samples containing ω -phase.

Electron microscopy was done after singlewave loading to 140kbar and multiwave loading to 200kbar at the initial temperature of 120°K. In both cases ω -phase was in the form of platelets 0,1-03 μ m thick (Fig. 3), which were fragmented and had higher defect concentration than α -phase. Analysis of diffraction patterns from the α and ω -crystals showed that both the $(001)_{\alpha}$ and $(011)_{\omega}$ planes, and the $[100]_{\alpha}$ and $[\bar{2}11]_{\omega}$ directions were parallel, which correlates with the orientation relationship between α and ω -phases in static-pressure treated titanium.⁽⁹⁾ The same results were obtained in the study of iodide titanium.

DISCUSSION

It is well known that the high velocity (close to that of sound) of the athermal martensite transformation is caused by cooperative lattice reconstruction.⁽¹⁰⁾ ω -phase formation during several microseconds at temperatures of 200-300°K allows one to assume the athermal martensite mechanism of $\alpha \rightarrow \omega$ transformation.⁽⁷⁾ Analysis of the lattice deformation for an α to ω -phase transition⁽⁸⁾ shows, in fact, that such reconstruction requires atomic displacements for the distances less than those of interatomic.

The present data about orientation relationships between the α and ω -phases in conjunction with the characteristic thin platelet form of ω -phase crystals, together with kinetic conditions of its formation, allows one to make the unambiguous conclusion that Ti $\alpha \rightarrow \omega$ transformation under shock loading proceeds by athermal martensite mechanism.

At the same time the investigation of $\alpha \rightarrow \omega$ transformation in titanium at static pressures shows rather slow isothermal isobaric kinetics.^(2,3)

In connection with essential differences between static and shock kinetics it is possible to plot another line (region) of athermal martensite $\alpha \rightarrow \omega$ transition, together with the well known