

The Effect of Thermomechanical Treatment Regimes on Microstructure and Mechanical Properties of V–Me(Cr, W)–Zr–C Alloys

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Abstract—The regularities of the formation of a heterophase structure in dispersion-strengthened vanadium V–Me(Cr, W)–Zr–C alloys are studied as a function of the regimes of their thermomechanical treatment. The regimes of treatment providing a substantial increase in the dispersity and homogeneity of spatial distribution of ZrC particles, temperature of recrystallization, and high-temperature (at $T = 800^\circ\text{C}$) short-time strength are found in comparison to conventional treatment regimes.

Keywords: vanadium alloys, thermomechanical treatment, electron microscopy, microstructure, dispersion hardening, mechanical properties.

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INTRODUCTION

In developing construction materials for the cores of thermonuclear and nuclear power reactors, great attention is paid to low-activation vanadium alloys of the system V–Ti–Cr (the basic alloy V–4Ti–4Cr [1–10]). Such attention given to vanadium alloys is due to their good functional properties and good processibility. Upon enrichment with interstitial elements (C, N, O) during melting and processing, vanadium alloys are typical heterophase dispersion-solidifying metallic systems with the possibility of further enhancing their functional properties using modifications of structural-phase states in the process of thermomechanical treatment (TMT) [7–10]. However, the conventional TMT regimes [3, 7] cannot provide a substantial increase in the characteristics of high-temperature strength of the alloys of the V–Ti–Cr system. This is due to the high inhomogeneity of the heterophase structure formed in the process of treatment of alloys and the formation of coarsely dispersed evolvments of oxycarbonitride phases being adverse factors leading to a decrease in the characteristics of the low-temperature plasticity and a decrease in the thermal stability of the microstructure and efficiency of dispersion hardening of alloys.

The possibility of substantially modifying the heterophase and imperfect structure of V–4Ti–4Cr alloys by transforming coarsely dispersed evolvments of interstitial phases into nanodimensional particles of

stable oxycarbonitride was shown in [8, 11, 12]. By virtue of this, new methods of TMT and chemical heat treatment (CHT) were developed [13–15] which provide preparing a distribution of particles of oxycarbonitride phases much more homogeneous over the bulk compared to conventional regimes of TMT [3, 7, 16], increasing the volume fraction of finely dispersed particles, and enhancing the thermal stability of microstructure and mechanical properties of alloys. However, the methods [8, 11, 12] based on a change in the mechanism of conversion of metastable vanadium carbides into particles of a stable oxycarbonitride phase do not provide a sufficient enhancement of high-temperature strength. A substantial disadvantage of the methods [14, 15] is the necessity to significantly complicate the technological treatment cycle of the items. In addition, the efficiency of dispersed strengthening of V–4Ti–4Cr alloys is substantially limited by relatively low values of the concentration of interstitial elements and, hence, a small volume fraction of particles of the second phase.

The problem of an additional enhancement of the high-temperature strength of vanadium alloys requires their additional doping and new methods of treatment to form in alloys special structural-phase states with a much higher (controllable) volume fraction of particles of strengthening nonmetallic phases and higher effects of dispersed hardening.

Table 1. Compositions of vanadium alloys (production of Bochvar High-Technology Research Institute of Inorganic Materials)

No.	Content of doping elements, wt % (at %)					
	Cr	W	Zr	C	O	N
Alloy 1 (weight of boule 0.9 kg)	8.75 (8.58)	0.14 (0.04)	1.17 (0.66)	0.01 (0.043)	0.02 (0.064)	0.01 (0.036)
Alloy 2 (weight of boule 1.2 kg)	4.23 (4.15)	7.56 (2.10)	1.69 (0.95)	0.02 (0.086)	0.02 (0.064)	0.01 (0.036)

In this work, we present the first results on the creation of vanadium alloys of the system V–Me(Cr, W)–Zr–C by their well-directed doping with interstitial elements (C, N, O) and those of substitution (W, Zr) and developing methods of modifying (nanostructuring) and enhancing the thermal stability of the heterophase structure of these alloys.

MATERIALS AND METHODS

The compositions of the alloys studied in this work, which were obtained at the Bochvar High-Technology Research Institute of Inorganic Materials, are presented in Table 1.

The initial samples of these alloys were sheets with a thickness of 1 and 3 mm and were obtained by the conventional TMT method (hereinafter, TMT-I), the main stages of which are presented below.

The thermal treatments of the alloys were carried out in shaft vacuum laboratory resistance furnaces of SShVL type in vacuum of $\sim 2 \times 10^{-5}$ Torr. In this case, the rate of cooling after thermal treatment in the temperature range of the secondary phase evolution changes in the course of cooling from ~ 5 (at $T = 1573$ – 1473 K) to 1 K/s (at $T = 1073$ K).

The structural studies were carried out using the techniques of optical metallography on Olympus GX-71 and NEOPHOT-21 microscopes and transmission electron microscopy on Philips CM-30 and SM-12 electron microscopes at an accelerating voltages of 300 and 120 kV, respectively.

The mechanical tests by the method of active elongation were carried out in vacuum of $\sim 2 \times 10^{-5}$ torr with a rate of deformation of $\dot{\epsilon} \sim 2 \times 10^{-3} \text{ s}^{-1}$ using samples in the form of double blades with the size of the working part of $15 \times 2 \times 0.8$ mm.

PROPERTIES OF STRUCTURAL-PHASE STATES AFTER CONVENTIONAL REGIME OF TMT-I

The conventional regime of TMT-I includes the following stages:

- a homogenizing vacuum annealing of a boule at a temperature of 1300°C for 8 h;
- an extrusion (pressing) at an elevated temperature;

- several cycles of rolling and setting at room temperature with intermediate vacuum annealings at temperature $T = 950$ – 1000°C ;

- a conclusive stabilizing vacuum annealing at 1000°C for 1 h.

The electron-microscopic study showed (Fig. 1) that, after such a treatment, the characteristic feature of the heterophase structure of both alloys under study is particles of the second phase with a size from ~ 0.1 to $0.3 \mu\text{m}$. The results of the electron-diffraction analysis indicate that they possess a face-centered cubic lattice with the parameter close to that of ZrC carbide. They can be identified as zirconium oxycarbonitride with a high content of carbon (hereinafter, as ZrC particles). Such particles hamper recrystallization of alloys (see Fig. 1), so that after TMT-I fine-crystalline or polygonal structure is formed in them (the size of grains and polygons is from ~ 1 to $\sim 3 \mu\text{m}$, Fig. 1a) with a high density of dislocations (about 10^{10} cm^{-2} , Fig. 1b).

The results of metallographic studies of the alloys after TMT-I and subsequent annealings at $T \leq 1400^\circ\text{C}$ (Fig. 2) showed that annealings for an hour at $T = 1200^\circ\text{C}$ result in the formation of nonequiaxial (inheriting the texture of rolling) grains of the primary recrystallization, indicating a rather intense develop-

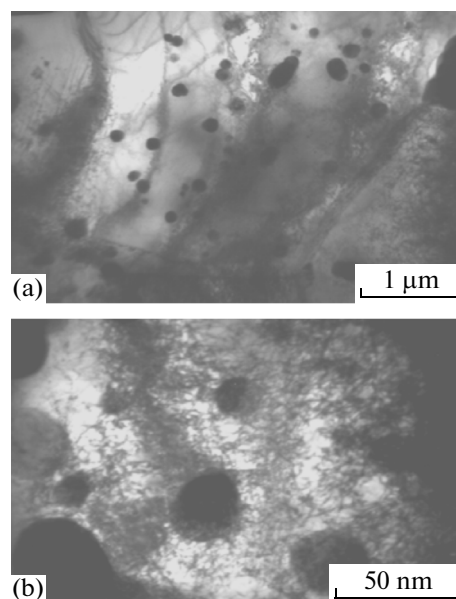


Fig. 1. Microstructure of vanadium alloys (see Table 1) after conventional regime of treatment (TMT-I): (a) alloy 1; (b) alloy 2.

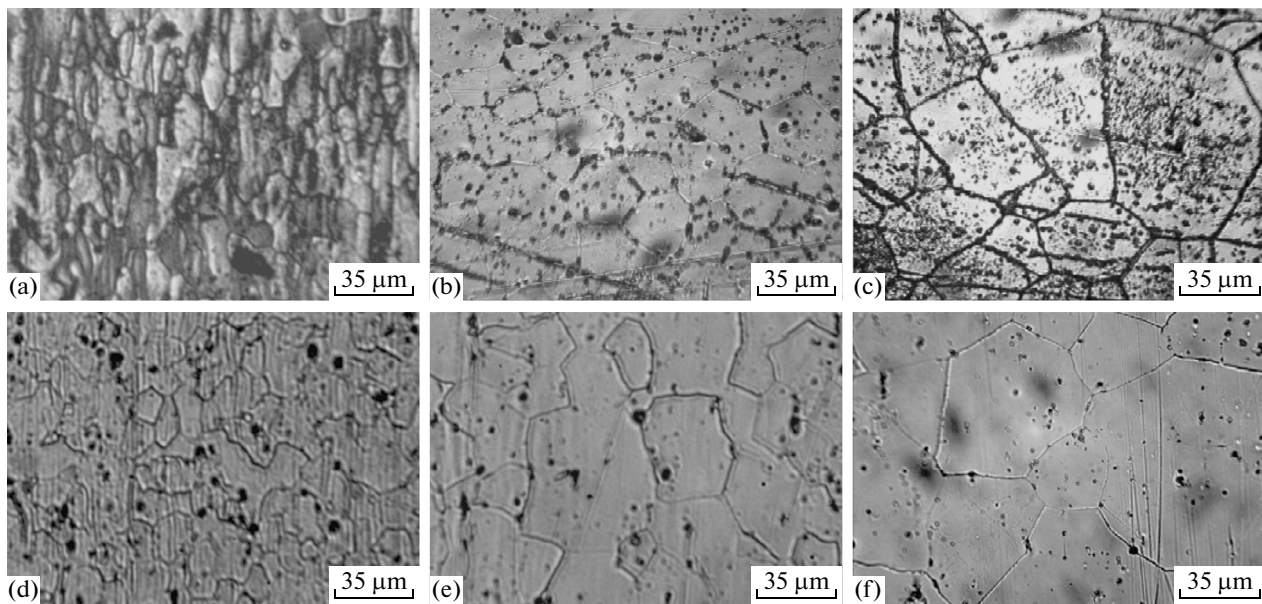


Fig. 2. Microstructure of alloys 1 (a, b, c) and 2 (d, e, f) after TMT-I and subsequent hour annealings at $T = 1200^{\circ}\text{C}$ (a, d), 1300°C (b, e), 1400°C (c, f).

ment of the processes responsible for returning to an imperfect substructure. The absence of any features of migration of intergrain boundaries is caused, evidently, by their fastening by ZrC particles stable at $T \leq 1200^{\circ}\text{C}$. These particles start intensely dissolving at $T \geq 1300^{\circ}\text{C}$, determining the temperature of the start of active collective recrystallization of alloys with the formation of equiaxial grains with sizes of about $30\ \mu\text{m}$ at $T = 1300^{\circ}\text{C}$ (see Figs. 2b and 2e) and greater than $50\ \mu\text{m}$ at $T \geq 1400^{\circ}\text{C}$ (see Figs. 2c and 2f). We note that, in alloy 2 after annealing at $T = 1300^{\circ}\text{C}$ (see Fig. 2e), a substantial number of rather fine grains (about $10\ \mu\text{m}$) of irregular shape remain, indicating substantial effects of fastening of their boundaries by particles of a nonmetallic phase at this temperature. Hence, the temperature of their complete dissolution or the temperature of this alloy's transfer into a single-phase state is $T \sim 1400^{\circ}\text{C}$.

The results cited indicate that the microstructure of the alloys studied here possesses much higher thermal stability (see Table 1) compared to the alloys of the V–4Ti–4Cr system. In reality, as was shown in [7, 8], in V–4Ti–4Cr alloys after TMT-I, the temperature of active collective recrystallization is $T_{\text{recr}} \sim 1000^{\circ}\text{C}$, which is 300°C lower than that in this work. This is probably the result of higher thermodynamic stability of zirconium carbide compared to the particles of stable oxycarbonitride TiV(C, N, O) in V–4Ti–4Cr alloys.

The numerous coarsely dispersed particles of the second phase detected in alloy 1 after annealings at $T \geq 1300^{\circ}\text{C}$ (see Figs. 2b and 2c), as in the case of similar thermal treatments of V–4Ti–4Cr alloys [7, 8], are particles of metastable vanadium carbide V_nC . They

evolve in the process of cooling of samples after annealings and their sizes are controlled by the diffusion of carbon atoms with very low energies of activation of migration of this element in vanadium [17] and the rate of cooling after thermal treatment. The known effect of lowering the barrier of nucleation of the second phases at defects of the microstructure [18] results in a maximum density of these particles at grain boundaries. After annealing at $T = 1400^{\circ}\text{C}$, a solid layer of metastable phase is formed at these boundaries. This is an extremely adverse feature of the microstructure, which negatively affects the physical and mechanical properties of the alloys and the possibility of modifying their heterophase structure in the process of various TMTs.

A comparison of the microstructure of alloys 1 and 2 after annealings at $T \geq 1300^{\circ}\text{C}$ indicates that doping with tungsten substantially decreases the size and density of metastable vanadium carbide particles both in the bulk and at grain boundaries. A similar effect found in [19, 20] in carbide-hardened niobium alloys is connected in [20] with a decrease in the rate of evolution of the particles mentioned caused by the necessity for tungsten possessing much higher energy of activation of migration compared to carbon to have a diffusion-controlled escape from the region of their formation.

REGULARITIES OF THE FORMATION OF MICROSTRUCTURE IN CONDITIONS OF NANOSTRUCTURING OF NONMETALLIC PHASE BY CHANGE IN THE MECHANISM OF $V_nC \rightarrow ZrC$ CONVERSION

As was shown by studying the regularities of phase transformations in the process of TMT for V–4Ti–

4Cr alloys [7, 8], an important factor of the formation of a heterophase structure of these alloys is the formation of particles of a stable oxycarbonitride $\text{TiV}(\text{C}, \text{N}, \text{O})$ from metastable vanadium carbides ($\text{V}_n\text{C} \rightarrow \text{TiV}(\text{C}, \text{N}, \text{O})$ conversion). Depending on the kinetic conditions of such a conversion, different mechanisms of it are possible:

(1) conversion in situ, when the reaction of the formation of a stable phase is implemented in place of coarsely dispersed metastable evolvments of V_nC and leads to an inhomogeneous coarsely dispersed heterophase structure;

(2) by dissolving metastable vanadium carbides with the subsequent evolving of the stable phase from the solid solution with the possibility of formation of a highly dispersed distribution of $\text{TiV}(\text{C}, \text{N}, \text{O})$ particles homogeneous over the bulk.

The formation in the process of a conventional regime of TMT-I of a high density of coarsely dispersed particles of the second phase is the result of the implementation of the first of the mechanisms mentioned of $\text{V}_n\text{C} \rightarrow \text{TiV}(\text{C}, \text{N}, \text{O})$ conversion. In addition, under the conditions of high inhomogeneity of the distribution of doping elements in a boule, coarsely dispersed particles of a stable phase can evolve according to this mechanism even at the stage of its homogenization.

On the strength of the above, in the process of developing new regimes of TMT which provide enhancement of the dispersity and homogeneity of the distribution of particles of nonmetallic phases, it is necessary to meet two conditions:

- to exclude the formation of coarsely dispersed particles of a stable nonmetallic phase in the process of homogenization of the boule;
- to create kinetic conditions for the implementation of the second of the aforementioned mechanisms of conversion of coarsely dispersed metastable phase V_nC into particles of stable oxycarbonitride $\text{TiV}(\text{C}, \text{N}, \text{O})$ in alloys V–4Ti–4Cr or ZrC in the alloys under study (see Table 1).

The first condition can be fulfilled by increasing the temperature of homogenization of the boule or by dissolving the coarsely dispersed particles of the stable oxycarbonitride that evolved at the stage of homogenization by thermal treatment of samples at a temperature exceeding the temperature of solubility of the secondary phases. As follows from the results presented, in this work for dissolving coarsely dispersed evolvments of ZrC in the process of the conventional regime of treatment of alloys (see Fig. 1), this temperature should be not lower than $T = 1400^\circ\text{C}$ (see Figs. 2e and 2f).

The regimes of the technological cycle of TMT providing the fulfillment of the second condition are selected from an analysis of the dependence on these regimes of the value of the kinetic parameter ($C_C D_C / C_{\text{Zr}} D_{\text{Zr}}$), controlling the mechanisms of con-

version of a metastable phase into a stable one. Here, C and D are the concentration and coefficient of diffusion of carbon (C) and zirconium (Zr) (an active carbon-forming element) in an alloy. As was shown in [8], the required values of this kinetic parameter are achieved by a simultaneous fulfillment of two conditions:

- a decrease in the temperature of intermediate annealings in the course of the technological cycle of rolling providing an increase in the ratio D_C / D_{Zr} ;
- a simultaneous substantial increase in the carbon concentration in the solid solution during conversion. In alloys with body-centered cubic lattices, this appears to be possible in structural states with a high density of dislocations.

Taking into account the above, for the alloys under studies (see Table 1), we have proposed a new TMT (TMT-II), in which after the third stage of TMT-I an hour annealing of samples at $T = 1400^\circ\text{C}$ is added, as well as several (not less than three) cycles of TMT “rolling $\varepsilon = 30\text{--}50\%$ at room temperature + annealing at $T = 600\text{--}700^\circ\text{C}$.”

The electron-microscopic study has shown that, after such a treatment in both alloys under study at a high density of particles of the second phase, which evolve at dislocations and have a size of several nanometers (Fig. 3a), coarsely dispersed evolvments of interstitial phases are absent.

Thus, the presented principles and methods of conversion of coarsely dispersed evolvments of interstitial phases into nanodimensional particles of nonmetallic phases are applicable both to V–4Ti–4Cr alloys containing interstitial elements (C, N, O) and to vanadium alloys with zirconium studied in this work.

The nanoparticles of ZrC type possess a relatively high thermal stability. Their high dispersity is retained in the process of thermal treatments at $T \leq 1200^\circ\text{C}$. As can be seen from Fig. 3b, after annealing at $T = 1200^\circ\text{C}$ (1 h), the size of most particles does not exceed 10 nm. By fastening individual dislocations and other elements of an imperfect substructure, these particles cause suppression of the collective recrystallization at this temperature. As after the conventional regime of thermomechanical treatment TMT-I, after TMT-II the collective recrystallization is observed in both alloys under study after hour annealings at $T \geq 1300^\circ\text{C}$. As is seen, this temperature does not depend on the TMT regime. This is caused, probably, by the fact that it is the temperature of the start of an intense dissolution and coagulation of ZrC particles, which is determined by the phase diagram of alloys and does not depend on such properties of the microstructure as dispersity and volume content of these particles. However, these properties appear to be the main ones upon a change in the mechanical properties.

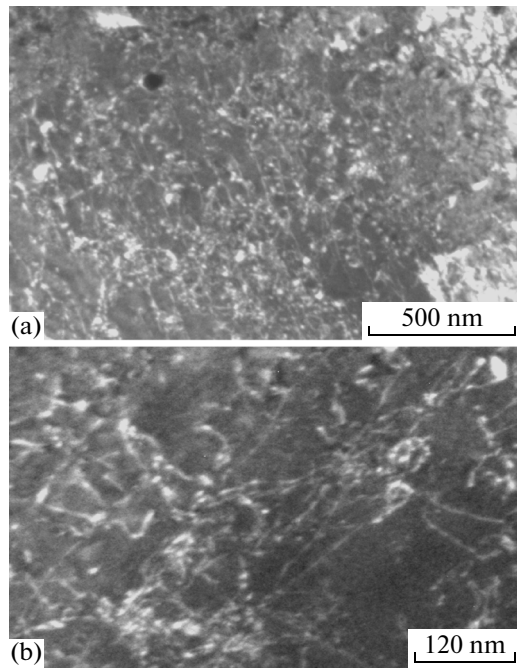


Fig. 3. Dark-field electron-microscopic images of ZrC nanoparticles and dislocation structure for alloy 2 after TMT-II (a) and subsequent annealing at $T = 1200^\circ\text{C}$ for 1 h (b).

INFLUENCE OF TMT REGIMES ON MECHANICAL PROPERTIES, PECULIARITIES OF PLASTIC DEFORMATION, AND DESTRUCTION

The results of the investigations into the influence of TMT regimes on mechanical properties of vanadium alloys, together with those obtained earlier [8, 11–13] by studying V–4Ti–4Cr alloys, are presented

in Table 2 and in Fig. 4. Of these results we will single out the following:

- After TMT-I (conventional regime of treatment), alloys 2 and 1 at $T = 800^\circ\text{C}$ possess strength characteristics close to V–4Ti–4Cr alloy but higher values of plasticity both at room and at an elevated temperature.

- Doping with tungsten results in a substantial (by ~25%) increase in the yield strength and ultimate stress at room temperature and weakly affects these characteristics at an elevated (800°C) temperature of deformation.

- Treatment according to the regime TMT-II increases by 30–40% the value of yield strength for alloys both at room temperature and at $T = 800^\circ\text{C}$. In alloy 1, these values exceed by ~20 MPa those after a similar regime of TMT in V–4Ti–4Cr alloy. In an alloy with tungsten (alloy 2), strengthening effects are more substantial and are comparable to those achieved in V–4Ti–4Cr alloy in the process TMT-III [12, 13], which includes large plastic deformations with the use of multiple confining forging.

- In alloy 1, the effects of strengthening in the process of deformation at room temperature are accompanied by a minor (from ~22 to 17%) change in the value of relative elongation to formation of a neck δ . In alloy 2 doped with tungsten (Fig. 4a), this value is practically unchanged.

The study carried out by scanning electron microscopy of the surface relief and fractography of samples of alloy 2 indicates the high viscosity of its destruction with the formation of a pit rupture (Fig. 5a).

A more intense decrease in the plasticity upon an increase in the strength is observed under the conditions of deformation at $T = 800^\circ\text{C}$. Nevertheless, in alloy 2, the level of plasticity at this temperature ($\delta \sim$

Table 2. Influence of TMT regimes on the mechanical properties of vanadium alloys

Regime of treatment	Temperature of tests $T = 20^\circ\text{C}$			Temperature of tests $T = 800^\circ\text{C}$		
	$\sigma_{0.1}$, MPa	σ_v , MPa	δ , %	$\sigma_{0.1}$, MPa	σ_v , MPa	δ , %
	Alloy 1 (V–Cr–Zr–C, N, O)					
TMT-I	240	395	22	180	235	22
TMT-II	330–350	490	17	240–250	290–300	4–5
	Alloy 2 (V–Cr–W–Zr–C, N, O)					
TMT-I	300	480	22	190	265	22
TMT-II	380	520–540	20–21	260–270	315–320	9–10
	Alloy V–4Ti–4Cr [8, 12, 13]					
TMT-I	290–300	450–460	19–21	170–180	310–350	18–20
TMT-II	320–330	480–520	16–18	220–230	380–400	9–12
TMT-III [12, 13]	383–398	530–550	18–20	260–280	440–470	8–10

$\sigma_{0.1}$ is the yield strength corresponding to the voltage at which a residual plastic deformation amounts to 0.1% of the length of the sample under test; δ is the value of relative elongation prior to formation of a neck; σ_v is the ultimate strength; TMT-III is a regime with the use of multiple confining forging [12, 13].

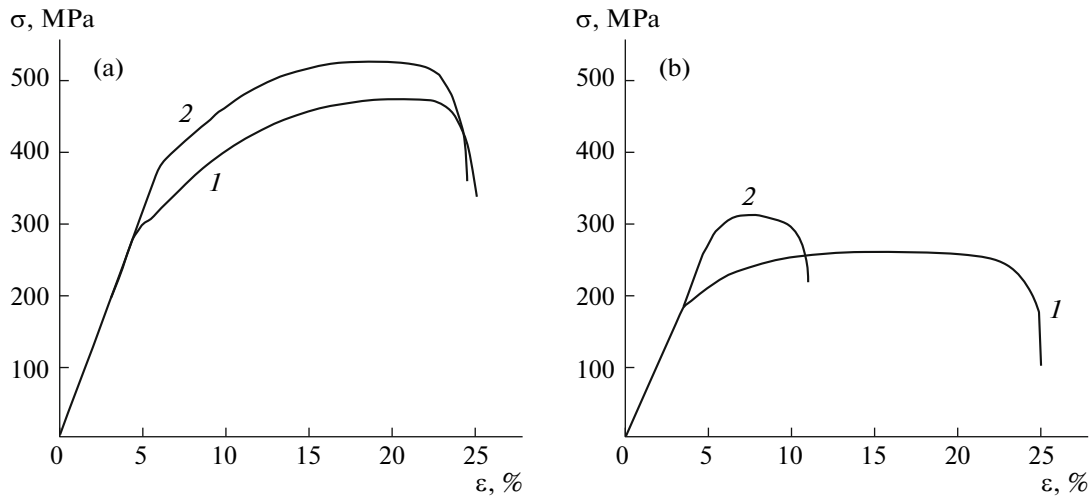


Fig. 4. Curves of deformation of alloy 2 at room temperature (a) and $T = 800^{\circ}\text{C}$ (b) after TMT-I (1) and TMT-II (2).

9%, Fig. 4b, curve 2), despite a 1.5-fold increase in the strength, remains rather high. The destruction at this temperature, like that at $T = 20^{\circ}\text{C}$, results in the formation of a pit rupture (see Fig. 5a), which indicates a rather high viscosity of destruction.

Another important feature of plastic deformation of alloys after TMT-II is its high homogeneity up to the formation of a neck. This is characteristic of deformation both at room temperature (see Fig. 5b) and at an elevated temperature ($T = 800^{\circ}\text{C}$, Fig. 6b).

This property is due to the high homogeneity of the structural state of alloys after TMT-II, in particular, to a conversion of coarsely dispersed evolvments into nanodimensional particles of the second phase homogeneously distributed over the bulk. At an elevated temperature, the high homogeneity of plastic deformation to a substantial degree is achieved owing to growth in the temperature of recrystallization of alloys after TMT-II and suppression the phenomenon of localization of deformation and destruction at grain boundaries. An example of such localization for the samples of V-4Ti-4Cr alloy recrystallized in the process of TMT-I is shown in Figs. 6c and 6d.

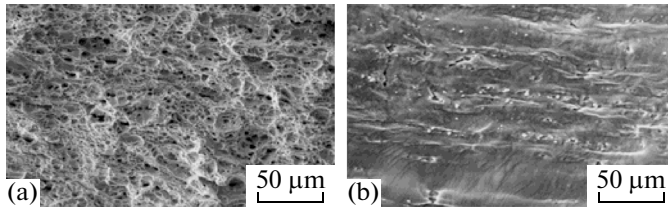


Fig. 5. The surface of destruction (a) and deformation relief of a side surface of the sample in the zone of formation of a neck (b) in alloy 2 after deformation at $T = 20^{\circ}\text{C}$ (scanning electron microscopy).

DISCUSSION OF RESULTS

The results presented indicate good prospects for the application of vanadium alloys of the V-Me(W, Cr)-Zr system (see Table 1) as more heat-resistant compared to alloys of the V-Ti-Cr system. Such prospects are due to the following:

- higher thermodynamic stability of ZrC carbides compared to oxycarbonitrides in V-4Ti-4Cr alloys, determining also higher temperature of recrystallization of alloys;
- the possibility of a cardinal modification of the heterophase structure (a substantial increase in the dispersity and homogeneity of the spatial distribution of particles of carbide phases) by methods converting coarsely dispersed evolvments of interstitial phases

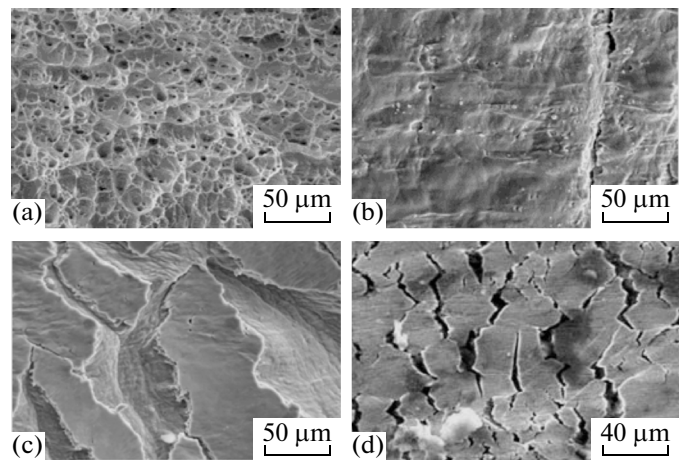


Fig. 6. The surface of destruction (a) and deformation relief of a side surface (b-d) of vanadium alloys in the zone of formation of a neck after deformation at $T = 800^{\circ}\text{C}$ (a-c) and 700°C (d) (scanning electron microscopy): (a, b) alloy 2 after TMT-II; (c, d) alloy V-4Ti-4Cr after TMT-I.

into nanodimensional particles of nonmetallic phases similar to the methods developed earlier [8, 11, 12] for alloys of the V–4Ti–4Cr system.

In the alloys studied (see Table 1), the developed regimes of TMT (TMT-II) provide a substantial (1.5-fold) increase in the high-temperature ($T = 800^\circ\text{C}$) short-time strength compared to the corresponding values by similar TMT regimes in V–4Ti–4Cr alloys. The higher, compared to oxycarbonitrides TiV(C, N, O) in V–4Ti–4Cr alloys, thermodynamic stability of ZrC particles formed in the process of TMT of Zr-containing alloys indicates good prospects for improving the characteristics of the high-temperature long-time strength in the alloys of the system V–Me(W, Cr)–Zr system. The effects of increasing the strength are achieved by keeping a sufficient reserve of plasticity.

Promising for further studies and development of new alloys and methods to increase their high-temperature strength by decreasing the susceptibility to cold brittleness are alloys of the V–Cr–W–Zr–(C, N, O) system. In these alloys, a substantial increase in the yield strength at $T = 800^\circ\text{C}$ is achieved practically without decreasing the high ($\delta \geq 20\%$) reserve of low-temperature plasticity observed at lower values of the strength of an alloy after treatment according to the regime of TMT-I (see Table 2). The properties mentioned can result from the following:

- A favorable influence of tungsten on the process of conversion of heterophase structure. This influence consists in a reduction of the rate of the formation and decrease in the sizes of coarsely dispersed particles of a metastable phase (vanadium carbide) in the process of cooling the samples at various stages of treatment of boules.

- The possibility of efficient solid-solution hardening with tungsten and the formation of its segregations at grain boundaries with a decrease in their susceptibility to grain-boundary sliding at an elevated temperature.

The possibilities for further increase in the high-temperature strength of vanadium alloys are associated with

- the necessity of applying compositions of new elements with variations (increase) in the volume fraction of nonmetallic phase by controllable change in compositions of alloys, first and foremost, with respect to interstitial elements (carbon and oxygen);

- developing and optimizing for these alloys TMT regimes providing the required relationship between the characteristics of high-temperature strength and low-temperature plasticity.

CONCLUSIONS

The results on the creation of vanadium alloys of the V–Me(Cr, W)–Zr–C system by their well-directed doping with interstitial elements (C, N, O) and substitutions (W, Zr) and the development of

methods for modifying (nanostructuring) and increasing the thermal stability of their heterophase structure are presented. The studied alloys of this system (see Table 1) are fabricated at the Bochvar High-Technology Research Institute of Inorganic Materials.

For carbide-hardened vanadium alloys V–Cr–Zr–C–N–O (alloy 1, see Table 1) and V–Cr–W–Zr–C–N–O (alloy 2, see Table 1), the regularities of the formation of a heterophase structure as a function of the regimes of their TMT are studied. The regime TMT-II is found, which provides

- kinetic conditions and the possibility of converting coarsely dispersed particles of metastable vanadium carbide into nanodimensional particles of a stable nonmetallic phase homogeneously distributed over the bulk based on zirconium carbide (ZrC);

- a relatively high thermal stability of a highly dispersed heterophase state and suppression of collective recrystallization to $T > 1200^\circ\text{C}$;

- high effects of dispersed hardening with an increase in a short-time strength while keeping a substantial reserve of low-temperature plasticity.

Alloys of the V–Cr–W–Zr–(C, N, O) system doped with tungsten are promising for further investigations and development of new vanadium alloys and the methods used to increase their high-temperature strength while decreasing the susceptibility to cold brittleness. In these alloys, a substantial increase in the short-time high-temperature strength is achieved while keeping the high low-temperature plasticity characteristic of the initial (prior to hardening) state (the value of a relative uniform elongation is more than 20%). The possibilities for a further increase in their high-temperature strength are associated with the variation (increase) in the volume fraction of the nonmetallic phase by a controllable change in the alloy composition with respect to carbon and oxygen.

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REFERENCES

1. M. I. Solonin, V. M. Chernov, V. A. Gorokhov, A. G. Ioltukhovskiy, A. K. Shikov, and A. I. Blokhin, *J. Nucl. Mater.* **283–287**, 1468 (2000).
2. R. J. Kurtz, K. Abe, V. M. Chernov, D. T. Hoelzer, H. Matsui, T. Muroga, and G. R. Odetter, *J. Nucl. Mater.* **329–333**, 47 (2004).
3. M. M. Potapenko, A. V. Vatulín, G. P. Vedernikov, I. N. Gubkin, V. A. Drobyshev, V. S. Zurabov, M. I. Solo-

- nin, V. M. Chernov, A. K. Shikov, I. P. Pazdnikov, and A. N. Rylov, *Vopr. At. Nauki Tekh., Ser. Materialoved. Nov. Mater.*, No. 1 (62), 152 (2004).
4. M. Koyama, K. Fukumoto, and H. Matsui, *J. Nucl. Mater.* **329–333**, 442 (2004).
 5. M. Hatekeyama, H. Watanabe, T. Muroga, and N. Yoshida, *J. Nucl. Mater.* **329–333**, 420 (2004).
 6. A. D. Korotaev, A. N. Tyumentsev, Yu. P. Pinzhin, and S. V. Ovchinnikov, *Vopr. At. Nauki Tekh., Ser. Materialoved. Nov. Mater.*, No. 1 (62), 163 (2004).
 7. A. N. Tyumentsev, A. D. Korotaev, Yu. P. Pinzhin, I. A. Ditenberg, S. V. Litovchenko, Ya. V. Shuba, N. V. Shevchenko, V. A. Drobyshev, M. M. Potapenko, and V. M. Chernov, *J. Nucl. Mater.* **329–333**, 429 (2004).
 8. A. N. Tyumentsev, A. D. Korotaev, Yu. P. Pinzhin, I. A. Ditenberg, V. A. Drobyshev, M. M. Potapenko, and V. M. Chernov, *Vopr. At. Nauki Tekh., Ser. Materialoved. Nov. Mater.*, No. 2 (63), 111 (2004).
 9. A. N. Tyumentsev, Yu. P. Pinzhin, S. V. Ovchinnikov, I. A. Ditenberg, A. D. Korotaev, Ya. V. Shuba, V. M. Chernov, and M. M. Potapenko, *Perspekt. Mater.*, No. 1, 5 (2006).
 10. L. M. Kryukova, M. M. Potapenko, V. M. Chernov, A. N. Ivanov, A. N. Tyumentsev, Yu. P. Pinzhin, and S. V. Ovchinnikov, *Vopr. At. Nauki Tekh., Ser. Materialoved. Nov. Mater.*, No. 1 (66), 152 (2006).
 11. Yu. P. Pinzhin, A. N. Tyumentsev, I. Yu. Litovchenko, S. V. Ovchinnikov, I. A. Ditenberg, N. V. Shevchenko, A. D. Korotaev, M. M. Potapenko, and V. M. Chernov, *Fiz. Mezomekh.* **7** (Spec. Iss., Pt. 2), 223 (2004).
 12. K. V. Grinyaev, I. A. Ditenberg, A. N. Tyumentsev, and V. M. Chernov, *Perspekt. Mater. (Spec. Iss.)*, No. 7, 89 (2009).
 13. A. N. Tyumentsev, A. D. Korotaev, Yu. P. Pinzhin, I. A. Ditenberg, S. V. Ovchinnikov, I. Yu. Litovchenko, V. M. Chernov, M. M. Potapenko, L. M. Kryukova, and V. A. Drobyshev, Patent No. 2360012, *Byull. Izobret.* No. 18 (2009).
 14. A. N. Tyumentsev, A. D. Korotaev, Yu. P. Pinzhin, I. A. Ditenberg, S. V. Ovchinnikov, I. Yu. Litovchenko, K. V. Grinyaev, V. M. Chernov, M. M. Potapenko, and V. A. Drobyshev, Patent No. 2445400, *Byull. Izobret.* No. 8 (2012).
 15. A. N. Tyumentsev, A. D. Korotaev, Yu. P. Pinzhin, I. A. Ditenberg, S. V. Ovchinnikov, I. Yu. Litovchenko, V. M. Chernov, M. M. Potapenko, and V. A. Drobyshev, Patent No. 2463377, *Byull. Izobret.* No. 2 (2012).
 16. M. M. Potapenko, A. K. Shikov, V. M. Chernov, G. P. Vedernikov, I. N. Gubkin, V. A. Drobyshev, and V. S. Zurabov, *Vopr. At. Nauki Tekh., Ser. Materialoved. Nov. Mater.*, No. 1 (64), 340 (2005).
 17. H. I. Chang, R. K. Viswanadham, and C. A. Wert, *Metal. Trans. A* **5**, 1907 (1974).
 18. K. V. Chuistov, *Aging of the Metallic Alloys* (Naukova Dumka, Kiev, 1985) [in Russian].
 19. F. Ostermann and F. Bollenrat, in *New Refractory Metallic Materials* (Mir, Moscow, 1971), pp. 130–155 [in Russian].
 20. F. Ostermann, *J. Less-Common Met.* **25**, 243 (1971).

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