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Microstructure and Mechanical Properties of Vanadium Alloys after Thermomechanical Treatments

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Abstract. The results of investigation of dispersion strengthening effect on parameters of structural-phase states and characteristics of short-term strength and ductility of vanadium alloys of V–4Ti–4Cr, V–2.4Zr–0.25C, V–1.2Zr–8.8Cr and V–1.7Zr–4.2Cr–7.6W systems with different concentration of interstitial elements after optimized thermomechanical treatment mode were summarized. It was shown that for effective realization of dispersion strengthening by Orowan-type mechanism at least 25–50% of the initial volume fraction of coarse particles should be transformed into fine-disperse state and redistributed over the volume of material.

INTRODUCTION

As it is well known [1–3], vanadium-based alloys are precipitation-hardened heterophase systems in which the formation of coarse (greater than 1 μm) precipitations of interstitial phases substantially limits their dispersion strengthening resources. As of today, it has become apparent that the use of traditional technological approaches in thermomechanical treatments (TMT) of these alloys is not capable of providing an effective modification of structural-phase state [4–6]. In [7–12] an optimized treatment mode (TMT II), allowing to carry out a modification of coarse phase to uniformly distributed fine (3–10 nm) particles in a controlled manner, was proposed. Such modification of heterophase structure in combination with grain microstructure transformation provides the most efficient dispersion strengthening implementation, leading to a substantial increase in short-term high-temperature strength maintaining acceptable levels of low-temperature plasticity.

This paper presents generalization and matching of the results of research of optimized TMT modes influence on parameters of structural-phase states and characteristics of short-term strength and ductility of vanadium alloys.

EXPERIMENTAL MATERIALS AND PROCEDURES

The study was conducted with the use of alloys of systems V–Ti–Cr (V–4.36Ti–4.21Cr–0.01C–0.02O–0.01N [7–9]), V–Zr–C (V–2.4Zr–0.25C–0.04O–0.01N [10, 11]) and V–Zr–Cr–W (V–1.17Zr–8.75Cr–0.14W–0.01C–0.02O–0.01N (1), V–1.69Zr–4.23Cr–7.56W–0.02C–0.02O–0.01N 2) 12]) (wt %) with different concentrations of interstitial elements after different TMT modes. The study of thin foils was carried out on transmission electron microscope Philips CM30 (300 kV). Research of specimens surface was conducted using a scanning electron-ion microscope Quanta 200 3D (30 kV). The applied methods of structural investigations, mechanical tests and specimens preparation methods are given in detail in [8, 11].

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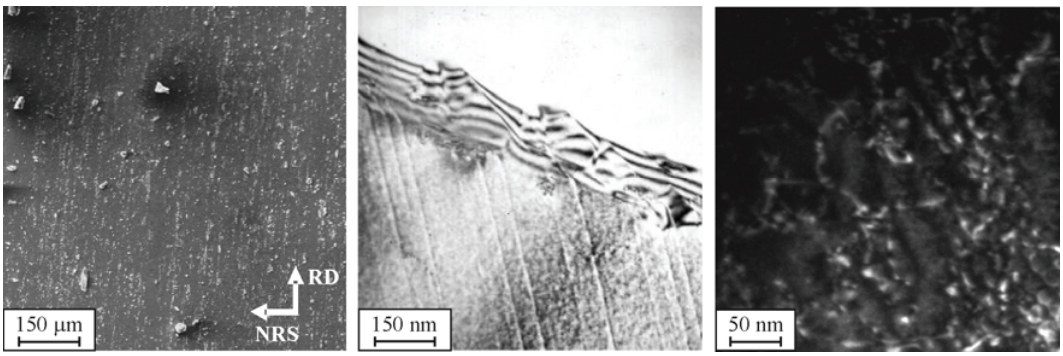

FIGURE 1. Typical structure states after TMT I. Scanning image of V–Zr–C alloy heterophase structure [10] (a), bright-field image of V–Ti–Cr alloy microstructure [9] (b), dark-field image of fine second phase particles in V–Zr–C alloy [10] (c). Scanning (a) and transmission (b, c) electron microscopy

RESULTS AND DISCUSSION

An important point is the fact that vanadium alloys used in current work are characterized by different volume fraction of the second phase, represented by nonmetallic compounds of complex oxycarbonitride type [8]. The results of evaluations, carried out on the basis of elemental composition data, showed that the entire volume fraction of second phase for the alloys of systems V–Ti–Cr is ~0.0025 [7, 9], V–Zr–C ~0.0236 [10, 11], V–Zr–Cr–W (1) ~0.0028 [12] and V–Zr–Cr–W (2) ~0.0038 [12].

Figure 1 shows microphotographs illustrating the effect of the traditional TMT mode (TMT I) on structural-phase state of vanadium alloys. As it can be seen (Fig. 1a), after rolling of alloys with a high content of second phase, stitched (in rolling direction) precipitation of coarse (5–10 μm) particles of these phases are frequently observed in cross-section. In addition, submicron-sized particles are observed in grains volume. Stresses, generated by the particles above, tend to result in cracking and delamination of the material. Presence of high density of sufficiently coarse particles at grain boundaries (Fig. 1b) adversely affects low-temperature plasticity (process-ability) of the alloys under study [8]. Fine particles of second phase are distributed over the volume in nonuniform manner and their observed volume fraction is commonly small (Fig. 1c).

Significant changes in the structural-phase state of investigated vanadium alloys after optimized treatment mode (TMT II) [7–12] can be clearly demonstrated by comparing Figs. 1 and 2.

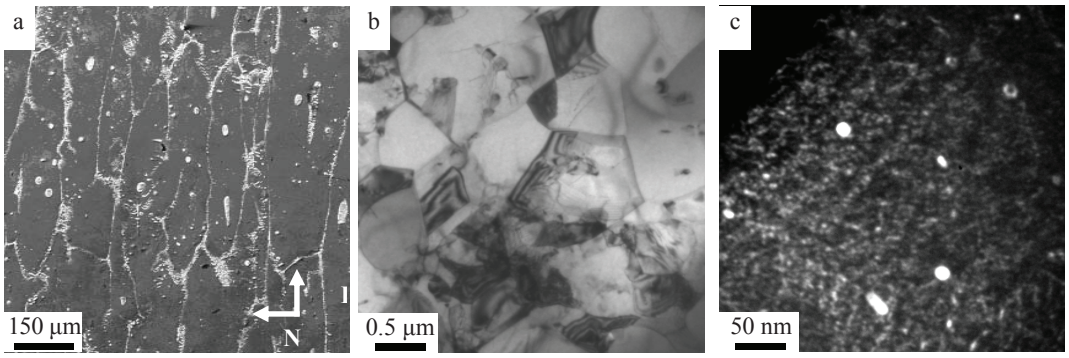

FIGURE 2. Typical structure states after TMT II. Scanning image of V–Zr–C alloy heterophase structure [11] (a), fine particles of second phase of V–Zr–C alloy [11] (b), bright-field image of V–Zr–C alloy microstructure [10] (c). Scanning (a) and transmission (b, c) electron microscopy

TABLE 1. Mechanical properties of vanadium alloys after different thermomechanical treatment modes.

Alloy	TMT mode	Testing temperature T = 20°C			Testing temperature T = 800°C		
		$\sigma_{0.1}$, MPa	σ_B, MPa	δ, %	$\sigma_{0.1}$, MPa	σ_B, MPa	δ, %
V–Zr–Cr–W (1) [12]	TMT I	240	395	25	180	235	25
	TMT II	330–350	490	20	240–250	290–300	7
V–Zr–Cr–W (2) [12]	TMT I	300	480	25	190	265	25
	TMT II	380	520–540	23	260–270	315–320	11
V–Zr–C [10, 11]	TMT I	280–300	380	16–20	170–190	260	14–17
	TMT II	370–400	490	14–17	250–280	310	14–18
V–Ti–Cr [8, 9]	TMT I	290–310	370	19–21	170–180	260	18–20
	TMT II	320–335	400	16–18	230–240	320	9–12
V–Ti–Cr [7]		370–380		23–24	270–280		13–15

$\sigma_{0.1}$—yield strength, σ_B—ultimate strength, δ—relative elongation up to necking

Scanning electron microscopic image of V–Zr–C alloy with high content of second phase demonstrates that after processing by TMT II instead coarse stitched structure (Fig. 1a) a mesh structure is formed (Fig. 2a), which is characterized by framing large areas of 100–500 μm in size by coarse particles. Meanwhile, many large particles are delaminated or ruined, forming conglomerates in local volumes. Within these framed areas fine-crystalline (grain size of 1–5 μm) or polygonal (polygon size of 0.5–1 μm) structural states are formed (Fig. 2b), fixed by fine-disperse (3–10 nm) particles of second phase uniformly distributed over the volume of material (Fig. 2c). It should be noted that in alloys with low volume fraction of second phase, there are no mesh configurations of particle precipitation, shown on (Fig. 2a), and the entire volume of material is presented by fine-crystalline or polygonal structural states like on (Fig. 2b).

In [8–12] it was demonstrated that TMT II application during modification of structural-phase state of different systems of vanadium alloys provides a significant (1.5–2 fold) increase in short-term high-temperature strength while maintaining an acceptable level of low-temperature plasticity (Table 1). Moreover, in [7] the possibility of effective application of modified treatment, based on TMT II principles, with rolling replaced by multi-directional forging, is demonstrated on the example of V–Ti–Cr alloy. As a result, short-time high-temperature strength is further increased by ~15% while maintaining high plasticity (Table 1).

In [7, 9–11] dispersion strengthening effect on strength properties of vanadium alloys of different systems was analyzed. The observed effect of strength increase (see Table 1) can be mainly attributed to the peculiarities of distribution of fine (a few nanometers in size) particles of second phase. In our opinion, significant $\Delta\sigma$ increase is due to particle size decrease while increasing their volume fraction, which ensures high effectiveness of strengthening by such fine particles of non-metallic phase in the process of overcoming them by glide dislocations by Orowan-type mechanism – rounding with elements of rapid climb [1, 2]. Achieving the uniform distribution of fine particles of minimum size over the volume provides maximum effectiveness of dispersion strengthening and contributes to the thermal stability of structural-phase state of all alloys studied in this work.

Estimates of dispersion strengthening effectiveness of the alloys under investigation are presented in Table 2, composed on the basis of data of works [7, 9–11]. Calculation of Orowan stresses values depending on the parameters of the heterophase structure—volume fraction and size of second phase particles—was carried by the formula (1):

$$\Delta\sigma = Gb/\lambda, \tag{1}$$

where $G \approx 47000$ MPa [13]—the shear modulus of vanadium; $b \approx 0.262$ nm—the Burgers vector of the dislocation; $\lambda \approx R(2\pi/3i)^{1/2}$—the distance between the particles; R—particle radius; f—their volume fraction.

As mentioned above, alloys under investigation are characterized by various volume fractions of interstitial element-based second phases. For example, for V–Zr–C-system alloys, volume fraction of particles f = 0.02 corresponds to the variant, when almost all of the carbide phase particles are in ultrafine state, f = 0.01—only half of these particles, etc. (Table 2). At the same time, for V–4Ti–4Cr-system alloys f = 0.0025 corresponds to almost the entire volume of the second phase, which is almost an order of magnitude smaller than in the alloy of V–Zr–C type (Table 2).

TABLE 2. Orowan stresses in vanadium alloys subject to volume fraction and sizes of second phase particles [7, 9–11]

Volume fraction, f	Particle size (diameter $2R$), nm				
	3	5	10	20	30
	Orowan stress $\Delta\sigma$, MPa				
0.02	790	510	226	120	80
0.01	560	360	160	85	56
0.005	400	240	120	60	40
0.0025	280	170	85	42	28
0.00125	201	120	60	30	20
0.000625	142	85	43	21	14
0.0003125	100	60	30	15	10

Analysis of Table 2 subject to volume fraction of the second phase of these alloys shows that only a portion of the total content of the initial coarse second phase precipitations is to be transformed into a fine state for effective implementation of dispersion strengthening.

SUMMARY

It is shown that the proposed thermomechanical treatment mode TMT II is a universal method of modification of the structural-phase state of a wide class of vanadium alloys with different elemental composition, type of hardening and volume content of the second phase. Dispersion strengthening effectiveness of the above alloys at the implementation of strengthening mechanism of Orowan-type mechanism was analyzed. According to the estimates, to increase the strength of the alloys ($\Delta\sigma$ up to ≈100 MPa), at least 25–50% of the initial volume fraction of coarse particles should be transformed into a fine state with a uniform distribution.

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REFERENCES

1. D. Korotaev, A. N. Tyumentsev, and V. F. Sukhovarov, *Dispersion Hardening of Refractory Metals* (Nauka, Novosibirsk, 1989).
2. V. K. Grigorovich and E. N. Sheftel, *Dispersion Hardening of Refractory Metals* (Nauka, Moscow, 1980).
3. V. I. Trefilov, Yu. V. Milman, and S. A. Firstov, *Physical Basis of Strength of Refractory Metals* (Naukova Dumka, Kiev, 1975).
4. T. Muroga, J. M. Chen, V. M. Chernov, R. J. Kurtz, and M. Le Flem, J. Nucl. Mater. **455**, 263 (2014).
5. H. Y. Fu, J. M. Chen, P. F. Zheng, T. Nagasaka, T. Muroga, Z. D. Li, S. Cui, and Z. Y. Xu, J. Nucl. Mater. **442**, S336 (2013).
6. P. F. Zheng, T. Nagasaka, T. Muroga, J. M. Chen, and Y. F. Li, Nucl. Eng. Des. **86**, 2561 (2011).
7. N. Tyumentsev, I. A. Ditenberg, K. V. Grinyaev, V. M. Chernov, and M. M. Potapenko, J. Nucl. Mater. **413**, 103 (2011).
8. N. Tyumentsev, A. D. Korotaev, Yu. P. Pinzhin, I. A. Ditenberg, S. V. Litovchenko, Ya. V. Shuba, N. V. Shevchenko, V. A. Drobishev, M. M. Potapenko, and V. M. Chernov, J. Nucl. Mater. **329–333**, 429 (2004).
a. I. A. Ditenberg, A. N. Tyumentsev, V. M. Chernov, and M. M. Potapenko, VANT Ser. Nucl. Fusion **2**, 28 (2011).
9. I. A. Ditenberg, K. V. Grinyaev, A. N. Tyumentsev, V. M. Chernov, and E. V. Chulkov, Russ. Phys. J. **55**, 223 (2012).
10. I. A. Ditenberg, A. N. Tyumentsev, K. V. Grinyaev, V. M. Chernov, M. M. Potapenko, and V. A. Drobyshev, Inorg. Mater. App. Res. **4**, 438 (2013).
11. N. Tyumentsev, I. A. Ditenberg, K. V. Grinyaev, I. V. Smirnov, V. M. Chernov, and M. M. Potapenko, *Physical Mesomechanics of Multilevel Systems-2014*, AIP Conference Proceedings 1623, edited by V. E. Panin, et al. (American Institute of Physics, Melville, NY, 2014), p. 643.
12. J. W. Martin, *Micromechanisms in Particle-Hardened Alloys* (Cambridge University, Cambridge, 1980).