

INVESTIGATION OF HEAT RESISTANCE AND STRUCTURE OF LAYERED REFRACTORY COMPOSITES REINFORCED BY CHEMICAL COMPOUNDS WITH SILICON, CARBON, AND BORON

ИССЛЕДОВАНИЕ ЖАРОПРОЧНОСТИ И СТРУКТУРЫ СЛОИСТЫХ ТУГОПЛАВКИХ КОМПОЗИТОВ, УПРОЧНЕННЫХ ХИМИЧЕСКИМИ СОЕДИНЕНИЯМИ С КРЕМНИЕМ, УГЛЕРОДОМ И БОРОМ

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Abstract: *In contrast to the melting technologies for the production of high-temperature alloys in the solid-phase technology developed by us, a multi-layer structure of the future composite material is already set at the initial stage of the billet formation. In the work presented, the billets were assembled from Nb- and Mo-foils with one or two-sided coatings of carbon or Si-B. Layered structures of composites were formed during the diffusion welding of packets at 1500°C and a pressure of ~14 MPa. In the place of the coatings, layers of high-temperature carbides or intermetallic Si and B compounds with Nb and Mo formed, which occupied up to ~40% of the cross-sectional area of the composites. The strength and crack resistance at room temperature, strength at a temperature of up to 1500 °C, as well as the modulus of elasticity and density were measured. The presence of intermetallides in the structure of composites leads to a decrease in their density and an increase in the modulus of elasticity.*

KEYWORDS: HEAT RESISTANCE, CRACK RESISTANCE, LAYERED COMPOSITE, LAYERED STRUCTURE, METAL MATRIX, SOLID-PHASE TECHNOLOGY, DIFFUSION WELDING, BENDING TESTS, INTERMETALLIC COMPOUND, MODULUS OF ELASTICITY.

1. Introduction

The result of many years of research on the heat resistance of alloys for the aviation industry, which began in the early 1970s, were alloys of the Ni-Al system. An important characteristic of a heat-resistant material is its operating temperature. The use of structural elements intended to work at elevated temperatures is always limited by the specific strength of the alloy from which they are made. This circumstance is due to the fact that an increase in the operating temperature with simultaneous increase in the mechanical characteristics of the alloy leads to a rapid expenditure of its strength capabilities.

Due to this, at present, heat-resistant (Ni-Al)-alloys, whose maximum melting point is slightly over 1600°C, began to fail to meet the ever-increasing demands of the industry. To date, they have reached their ceiling of operating temperatures equal to of ~1150°C, and they need an effective replacement.

Arising from this flow of information on heat-resistant materials instead of Ni-Al system produced, mainly smelting technology continues unabated. Most of the research fall on niobium and molybdenum alloys with silicon and, to a lesser extent, on Nb-Al alloys with intermetallic hardening. But the melting method for the manufacture of products from heat-resistant alloys, in particular gas turbine engine blades, is precision investment casting, which was developed for the Ni-Al system. For high-temperature alloys Nb-Si and Mo-Si practical realization of such a casting is associated with the need to develop new refractory and inert with respect to melt the ceramic materials for the crucibles.

We have proposed and for almost a decade has been developing a solid-phase technology for obtaining composite heat-resistant materials of a layered type. It differs from traditional methods of melting, for example, the method of directional crystallization, in that the formation of the desired structure occurs at the solid-phase interaction of layers, and the multilayeredness and orientation of the structure is created at the initial stage of the assembly of the packet. Heat resistance is provided by the creation of 2-dimensional planar layered structures of intermetallic compounds or carbides, and high crack resistance by a metal matrix, which often has its own layered structure. We call it a second-order structure. Solid-phase methods of obtaining heat-resistant composite materials have recently shown increasing interest.

The essence of solid-state technology can be understood from the following description. Layered composites consisting of

alternating layers of intermetallic compounds and layers of Al solid solution in niobium Nb(Al) were obtained as a result of diffusion interaction between Nb- and Al-foils during the diffusion welding of Nb/Al packets using pressure [1, 3]. As in our studies [3-5], the phase composition of the interaction zone between niobium and aluminum depended on the ratio t_{Nb}/t_{Al} , where t_{Nb} and t_{Al} are the foil thicknesses of niobium and aluminum, respectively, and the final welding temperature. After 1500°C in a composite with $t_{Nb}/t_{Al} = 1.4$, a structure was formed from Nb(Al), Nb₂Al, and Nb₃Al.

The Nb-Al system is most simple for the use of solid-state technology in that aluminum is a deformable metal. This makes it easy to form the original multilayered packets from Nb- and Al-foils. To study the heat resistant layered composites with silicides, carbides and borides during assembly of source packets used metallic ribbons, respectively with coatings of silicon, carbon and boron or mixtures thereof, applied by various methods.

The relevance of this work is to obtain and investigate the heat-resistant material, which we believe can operate at higher temperatures than not only the currently widely used (Ni-Al)-alloys, but also alloys based on the Nb-Al and Nb-Si systems with intermetallic hardening. As follows from our studies [6-8], alloys of niobium with a directed eutectic (Nb)-Nb₅Si₃ at temperatures above 1300°C begin to degrade.

To achieve a higher temperature level of heat resistance than that of the Nb-Si alloy, it is proposed to start by alloying its silicide's with carbon and boron. For example, the NbB₂ boride melts only at ~3000°C. This is almost 500°C higher than the melting point of the silicide Nb₅Si₃ and exceeds the melting point of Nb₃Al by more than 900°C. Similar studies have been started with the Mo-Si system. It was shown in [9] that the Mo-Si-B ternary alloys possessed good heat-resistant properties at high temperatures.

The interaction of the components and the formation of the orientation of the structure occurred in multilayered packets composed of Nb and Mo foils with (Si-B) or (Si-C) coatings, during their diffusion pressure welding. Multilayered composites of niobium with carbide hardening were also studied. In this case, multilayer packets of Nb-foils with a pyrolytic coating were welded.

2. Composites Nb-Si-B u-Mo-Si-B

Packets assembly. The Nb/(Si-B)- and Mo/(Si-B)-packets were assembled from Nb- and Mo-foils with an (Si-B) coating (Table 1). Suspensions from a mixture of Si- and B-powders with a Si/B ratio of 3 in polyethylene glycol were applied to the foils by the

"staining" method. The width and length of the packets are 35 and 40 mm, respectively. The direction of the rolling foils – along the length of the package.

Diffusion welding (DW). Welding of packets was carried out in a vacuum (not lower than 10^{-4} mm Hg) installation with a graphite heater designed for a maximum temperature of 1700°C and a maximum load of 10 T. Polyethylene glycol was evaporated, starting from ~ 200°C, as the packet was heated.

Table 1. Description of packets and modes of diffusion welding

Package description and DW modes	Quantity, pcs.
Mo/(Si-B)-1 packet	
Mo-foil 50 μm thick with 2-sided (Si-B) coating	15
Mo-foil 50 μm thick with 1-sided (Si-B) coating	2
Total number of Mo-foils	17
The total number of double (Si-B)-coverings	16
Weight of all Mo-foils without coating – 10.370 g	
The weight of the packet after DW is 10.880 g	
The weight of all (Si-B) coatings is 0.510 g or 4.7% by weight	
The thickness of the packet after DW is 1.10 mm	
DW: 1500°C, 30 min at 10.5 MPa (1.5 T) + 30 min at 14.0 MPa (2 T)	
Nb/(Si-B) packet	
Nb-foil 60 μm thick with 1-sided (Si-B) coating	21
Nb-foil 60 μm thick without (Si-B) coating	1
The total number of Nb-foils	22
The total number of (Si-B)-coverings	21
The weight of the packet after DW is 17.315 g	
The thickness of the packet after DS is 1.7 mm	
DC: 1500°C, 1 ч при 14.0 МПа (2 Т)	

Layered structure studies

Composites of Mo/(Si-B). The macrostructure of the cross-section of the Mo/(Si-B)-1 composite (Figure 1) after diffusion welding consisted of alternating layers of a light background and an interaction zone (IZ) between molybdenum and an (Si-B) coating. IZ were uneven in thickness – from their total absence to ~ 70 μm . At a thickness of 40-70 μm (Fig. 2a), it has a large number of Kirkendal pores inside. In pores of 20-30 μm thick, the pores were kept in an insignificant amount (Fig. 2b). It would seem that with an even smaller thickness, the number of pores would have to decrease and result in their total absence. But, starting from a thickness of ~ 20 μm , both the absence of pores and the existence of voids with a length of up to several hundred microns were observed. And so down to the complete absence of the diffusion layer. Such an ambiguity in the change in the structure of IZ at their small thicknesses can be explained by the fact that such voids can form in the place of a fragile and not strong intermediate phase and, therefore, tend to locally crumble at the preparation of a metallographic section for microscopy.

In the Mo-layers near and at a certain distance from the IZ, the boron content varied within the limits of 17-28 at.% and almost no silicon. However, according to their state diagrams with molybdenum, at 1500°C the solubility of boron in Mo is very small, and silicon can be dissolved up to ~ 2 at.%. The compositions obtained correspond to the eutectic region Mo(B) + Mo₂B of the Mo-B diagram, where Mo(B) is a solid solution of boron in Mo.

Phase structure of interaction zones. Their extreme layers of thickness ~ 10 μm bordering the Mo(B) + Mo₂B layers consisted of molybdenum boride Mo₂(B,Si) with silicon dissolved in it. The averaged Mo, B, and Si contents in them are Mo_{1.87}(B_{0.99}Si_{0.14})_{1.13} and Mo_{1.74}(B_{1.22}Si_{0.04})_{1.26} for the IZ in Fig. 2a and b, respectively. That is, in the boride Mo₂B, which is located in the IZ with a small amount of pores, Si is much less dissolved.

In the IZ with a porous middle between it and Mo₂(B,Si), there were 2-phase regions 20-25 μm thick (see Fig. 2a) consisting of intermetallic compounds Mo(B,Si) \equiv Mo_{1.06}(B_{0.63}Si_{0.31})_{0.94} and Mo₅(Si,B)₃ \equiv Mo_{4.29}(Si_{2.42}B_{1.29})_{3.71} or solid solutions of Si in α MoB

and boron in the silicide Mo₅Si₃, respectively. An analogous phase composition had the middle region of IZ in Fig. 2b.

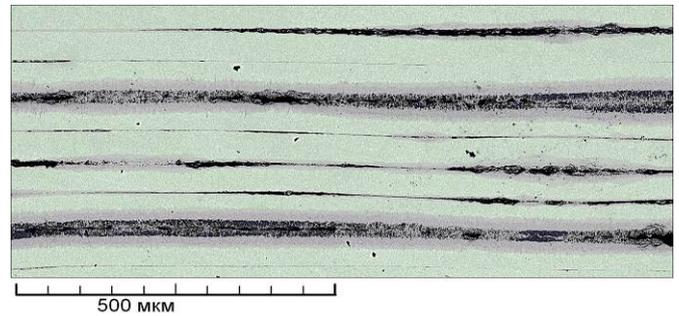


Fig. 1. The macrostructure of the cross-section of the Mo/(Si-B)-1 composite, which coincides with the Mo-foils rolling plane

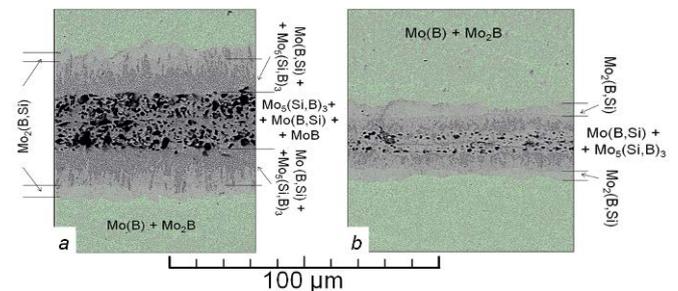


Fig. 2. Composite Mo/(Si-B)-1. The microstructure of the interaction zones between Mo and the coating with porous middle (a) and with a small number of pores (b)

The middle of pollutants with a large number of pores (see Fig. 2a) consisted of two phases: Mo₅(Si,B)₃ and Mo(B,Si). The Si and B concentrations in the Mo₅(Si,B)₃ compound varied in favor of boron from 35.0 Si and 3.4 at.%B to 23.1 Si and 14.4 at.% B or from Mo_{4.93}(Si_{2.80}B_{0.27})_{3.07} to Mo_{5.00}(Si_{1.85}B_{1.15})_{3.00}. A similar composition change was observed in the Mo(B,Si) boride: from 37.9 B and 19.2 at.%Si to 42.0 Mo + 57.8 at.%B or from Mo_{0.86}(B_{0.76}Si_{0.38})_{1.14} to Mo_{0.84}B_{1.16} (+ 0.13 at.%Si).

Composite Nb/(Si-B). Unlike Mo/(Si-B), the cross-sectional structure of a composite with Nb was regular (Fig. 3a). Zones of interaction (IZ) formed in situ (Si-B)-coatings occupying ~ 40% of the cross-sectional area were alternated with bright layers of boron solid solution in niobium Nb(B): from ~ 0 to 3.85 at.%B.

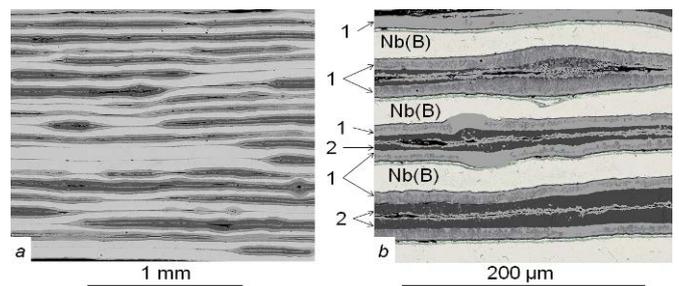


Fig. 3. Macro- (a) and microstructure (b) of the cross-section of the Nb/(Si-B) composite after DW in a plane parallel to the rolling direction of Nb-foils

We will trace the phase composition of the IZ with the help of Fig. 3b. Two of its outer layers of gray (1) bordering with Nb(B) consisted of the intermetallic compound Nb₅(Si,B)₃, as the main (matrix) phase and inclusions of Nb(B,Si). The composition of Nb₅(Si,B)₃ varied from the silicide Nb₅Si₃ (boron-free) to Nb_{5.24}(Si_{1.76}B_{1.00})_{2.76} with 12.5 at.% boron. The composition of the inclusions corresponded to the formula Nb_{1.03}(B_{0.74}Si_{0.23})_{0.97}.

The thin inner layer and the intermediate layers of dark color (2) were single-phase and consisted of ternary intermetallic compounds Nb(Si,B) \equiv Nb_{1.01}(Si_{0.59}B_{0.40})_{0.99} and Nb(Si,B)₂ \equiv Nb_{0.92}(Si_{1.59}B_{0.49})_{2.08}, respectively.

Characteristics of high-temperature strength of composites.

The fracture toughness evaluation was carried out on samples with a lateral notch, and to calculate the critical factor of stress intensity K^* under the conditions of plane deformation, the formula [10] was used: $K^* = (PS/BW^{3/2}) \cdot f(a/W)$, where P is the maximum load, S is the distance between the supports, B and W are the thickness and height of the sample, and a is the value of the lateral incision along the vertical. Tests of Nb/(Si-B) composites on crack resistance and strength at room temperature were carried out at a perpendicular (\perp) and parallel (\parallel) application of the load with respect to the surface of the layers.

Anisotropy of the structure of the material, led to a difference in the values of K^* , as well as the variance of their values. The following fracture toughness values were obtained: $K^*_{\perp} = (13,0 \pm 1,3)$ and $K^*_{\parallel} = (12,0 \pm 0,5) \text{ MPa} \cdot \text{m}^{1/2}$. The values of K^* show that the material occupies an intermediate position between traditional ceramics and high-strength metal alloys.

The strengths σ_B when the load is applied in directions perpendicular and parallel to the plane of the layers are $\sigma_{B\perp} = (600 \pm 180) \text{ MPa}$ and $\sigma_{B\parallel} = (790 \pm 48) \text{ MPa}$, respectively. The higher strength with parallel layer load can be explained by the higher moment of resistance at bending of high modulus and high strength "boridosilicidal" layers than with the load perpendicular to the layers.

In Fig. 4 the results of testing for the composites Nb/(Si-B) and Mo/(Si-B) at high temperatures in the range 1100-1500°C are shown. The dependence $\sigma_B(T)$ draws attention to the fact that the values of the strength of about 250 MPa, which are satisfactory for such temperatures, remain in the entire high-temperature interval.

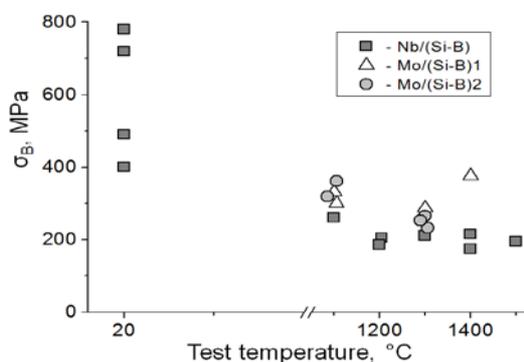


Fig. 4. Dependence of the strength σ_B of the composite samples Nb/(Si-B) and Mo/(Si-B) on the test temperature at a perpendicular load application with respect to the surface of the layers

3. Composites with carbide-silicide hardening

1. A composite obtained from a packet with 30 Nb-foils having a one-sided suspension coating from a mixture of Nb-Ti-Cr-Mo-Si-Al-ZrH₂ powders was studied. Welding of the packet and solid phase formation of the layered structure occurred at 1500°C under a pressure of 8.4 MPa for 5 hours in a CO atmosphere. The macro- and microstructure of the cross section of the packet is shown in Fig. 5.

Three strengthening phases are distinguished, different in composition and color contrast: 1) TiC carbide with 47.3 ± 0.8 Ti and 51.2 ± 1.1 at.% C, in which are dissolved from 0.4 to 1.1 Nb and from 0.4 to 0.7 at.% Zr; (2) carbidosilicide $\text{Me}_{0.975}(\text{C}_{0.51}\text{Si}_{0.49})_{1.025}$ containing 29.5 ± 5.4 Ti, 11.7 ± 4.6 Nb, 5.4 ± 0.6 Zr, 1.3 ± 0.1 Al, 0.7 ± 0.5 Mo and ~ 0.2 at.% Cr; (3) Nb₅Si₃ intermetallide containing 53.7 Nb, 30.6 at.% Si, and 15.0 at.% C, the balance being 0.45% Ti and 0.2 at.% Cr; (4) pores formed during the grinding of the sample, in places where the brittle phase is deposited.

The remaining volume was a layered viscous-plastic matrix whose composition varied from Nb-solid solution with ~ 99.5 wt.%Nb to (Nb-Ti)-alloy with 67–76% Nb and 20–15 wt.%Ti. Significant concentrations have: Mo – from 1.1 to 5.6%, Cr – from 1.7 to 2.0% and Al – from 0.8 to 1.1% by weight. In addition, the alloy contains silicon – from 0.3 to 0.4 wt.% and a sufficient

amount of carbon from 2.8 to 4.8 wt.% (15.3 – 24.4 at.%). And also in the matrix there are inclusions corresponding to carbide Me₂C, where Me = Nb (52.8%), Ti (8.85%), Cr (1.15%) and Mo (0.9 at.%). Carbon is contained in an amount of 34.9 at.%.

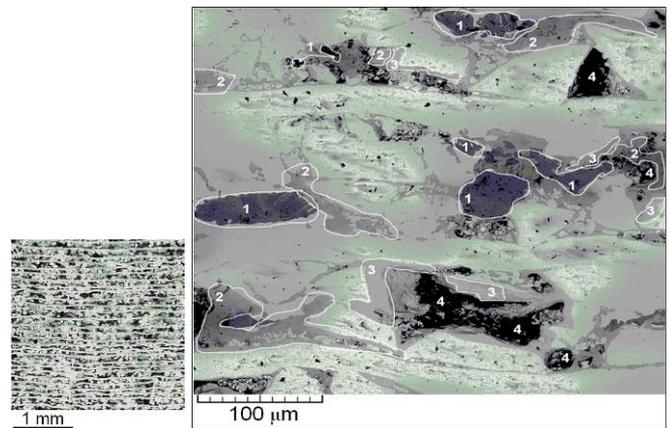


Fig. 5. Macro- (left) and microstructure of the composite (right): 1, 2 and 3 - strengthening phases; 4 - pores

Mechanical tests using the 3-point bending method at 1100-1300°C showed the following results: the stretch of proportionality $\sigma_{PR} = 200\text{--}370 \text{ MPa}$, $\sigma_B = 260\text{--}390 \text{ MPa}$. At room temperature, $\sigma_B = 320\text{--}540 \text{ MPa}$.

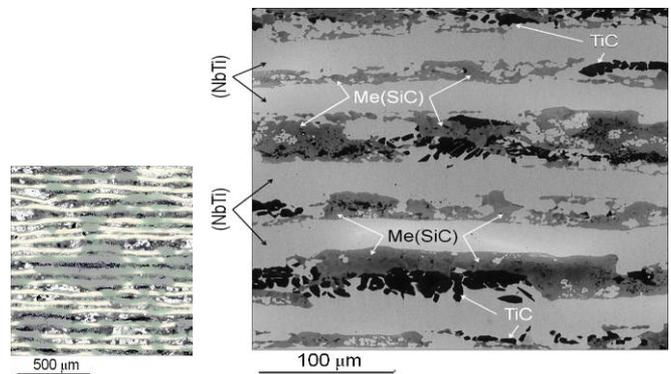


Fig. 6. Macro- and microstructures of the composite after welding at 1500°C and 14 MPa pressure for 3 hours in an atmosphere of CO

2. Composite samples 2–4 thick and $\sim 30\text{--}40 \times 60 \text{ mm}$ in size were prepared by welding of multilayered packets composed of 30 Nb-foils 30 μm thick with a coating similar to the previous one, but without Nb: 49.1Ti–19.8Mo–12.4Si–12.1ZrH₂–3.8Cr–2.8 wt.% Al.

The multilayer structure (Fig. 6) consisted of layers of a multicomponent alloy (Nb, Ti) of thickness $\sim 30 \mu\text{m}$ alternating with interlayers of approximately the same thickness from the strengthening phases. The following phases are identified.

1. TiC carbide with 50.4Ti and 47.7 at.%C, in which 1.0%Nb and 0.6 at.%Zr are dissolved.

2. Carbo-silicide $\text{Me}(\text{Si,C}) \equiv \text{Me}_{1.05}(\text{Si}_{0.51}\text{C}_{0.44})_{0.95}$ with content of the elements: 27.8Ti–14.5Nb–6.9Zr–2.5Al–0.6Mo–0.3 at.% Cr and 47.4 at.%(C + Si).

3. Viscous-plastic structural component (Nb,Ti), the composition of which (based on the Nb–C diagram) varied from the hypereutectic region $\beta\text{-Nb}_2\text{C} + (\text{Nb})$ with 33.4 Nb and 30.1 at.% Ti to the Nb-solid solution with 91.6Nb and 5.6 at.%Ti. The content of C varied from 22.1 ± 2.7 to ~ 0.2 at.%.

4. Niobium composites with carbide hardening

Before the assembly of the Nb/CNbC/CNbC/... packets, CNbC foils were prepared. These are Nb-foils with a 2-sided pyrolytic coating deposited in one experiment on both surfaces of niobium at 1100°C by decomposition of methane CH₄. Two modes of DW packets were used: 1500°C for 30 min and 2 h at a pressure of 8.2 and 32.7 MPa, respectively.

The light layers (Nb) in both composites were an Nb-solid solution with a carbon content of 1.2 ± 0.6 and 0.5 ± 0.4 wt.% (Figures 7a and b, respectively). Solid gray layers in the composite 1CNbC/CNbC (a) - Nb₄C_{3-x} carbide with a concentration of 37.2 ± 1.6 at.%C. The same layers in the 2nd composite (b) contained 24.7 ± 1.1 at.%C and consisted of Nb₂C carbide.

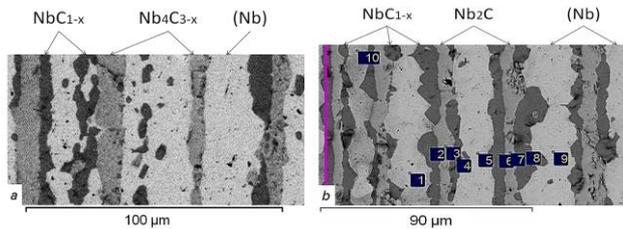


Fig. 7. Microstructure of the composites 1CNbC/CNbC and 2CNbC/CNbC after DW for 30 min under a pressure of ~8 MPa (a) and 2 h under a pressure of ~33 MPa (b), respectively

Discrete inclusions of dark color located along the boundaries of continuous carbide layers in 1CNbC/CNbC and 2CNbC/CNbC composites contained 61.5 ± 3.5 and 54.8 ± 1.5 at.% C, respectively, which is more than 50 at.%, and exceeds the boundaries of the NbC_{1-x} existence region. But there are no other niobium carbides other than NbC_{1-x} with a carbon content exceeding the C content in Nb₄C_{3-x}. Therefore, we believe that dark inclusions are nothing more than carbide NbC_{1-x}.

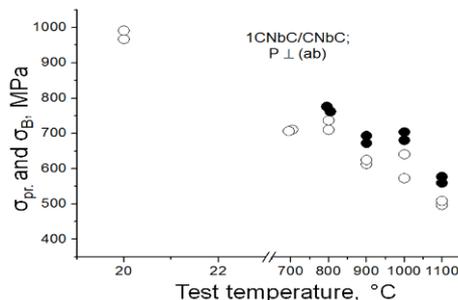


Fig. 8. Dependences of σ_{PR} (○) and σ_B (●) on the test temperature for the 1CNbC/CNbC composite at $P \perp$ (ab)

The results of mechanical tests at temperatures up to 1100°C and a load P perpendicular to the layers were carried out for a composite with DW for 30 min. Voltage σ_B in the temperature range from 700 to 1100°C dropped ~750 to ~550 MPa.

5. Conclusion

1. By solid-state interaction in process of diffusion welding of multilayer packets composed of thin Mo- and Nb-foils with suspension (Si-B) coatings layered composites containing intermetallic of metals with B and Si as potential hardening phases were obtained. Room strength flexural tests for perpendicular (\perp) and parallel (\parallel) orientations of load application and rolling direction of Nb-foil showed $\sigma_{B\perp} = (600 \pm 180)$ and $\sigma_{B\parallel} = (790 \pm 48)$ MPa. A significant variance for $\sigma_{B\perp}$ leveled the expected difference in strength values $\sigma_{B\perp}$ and $\sigma_{B\parallel}$. High-temperature tests of Nb/(Si-B) and Mo/(Si-B) composites in the interval 1100-1500°C showed a satisfactory result for such temperatures – the strength of $\sigma_{B\perp}$ of the order of 250 MPa, which was preserved throughout the temperature range. The tests for crack resistance showed that the material occupies an intermediate position between ceramics and high-strength metal alloys. But composites have the potential to increase it by optimizing the microstructure of the interaction zones, varying the parameters of DW. The studies were of a "pilot" nature. Both composites have reserves for increasing fracture toughness. Studies on creep resistance at high temperatures and optimization of the initial composition and regimes of DW are also necessary. This determines the direction of further research.

2. With the example of composites prepared by welding of multilayer packets made up of Nb-foils with suspension

multicomponent coatings, it has been shown that by solid-state technology can produce layered composites with structures of a viscous-plastic matrix and high-temperature carbide and carbide-silicide phases. Carbon contained only in the atmosphere of the chamber in the form of CO at a residual pressure of $\sim 10^{-4}$ mm Hg. Due to its atomic mobility, it penetrated into the structural components of the composites. With the participation of carbon, strengthening phases in the form of (Ti-Nb)-carbide and layers of carbide-silicide compounds Me(C,Si) were formed.

3. A composite with a layered structure hardened by carbide layers was prepared by the DW under the pressure of multilayer packets of Nb-foils with a 2-sided pyro-carbon coating. Its structure was characterized by the alternation of the layers of the solid solution C in Nb and the layers consisting of NbC_{1-x} and Nb₂C or NbC_{1-x} and Nb₄C_{3-x} carbides. The stretch of proportionality of PC at room temperature of such composites was 935-980 MPa. For a multilayer Nb/Nb composite, the $\sigma_{PR} = 290-300$ MPa. At 1100°C, a layered composite containing 33.6% by volume of carbide layers had σ_{PR} of 495-510 MPa.

References

1. Chung Dong-Seok, Enoki Manabu, Kishi Teruo. Microstructural analysis and mechanical properties of in situ Nb/Nb-aluminide layered materials // Science and Technology of Advanced Materials, 2002, v.3, No 2, p. 129–135.
2. Tae Ahn Byung, Hyeontag Jeon, Young Hur Bo, Kibae Kim, Wan Park Jong. Fabrication of metal/intermetallic compound laminate composites by thin foil hot press process // Solid State Phenomena, 2007, v. 124–126, p. 1099–1102.
3. Коржов В.П., Карпов М.И., Прохоров Д.В. Многослойная структура и высокотемпературная прочность жаропрочных материалов на основе соединений ниобия с алюминием и кремнием, полученных из композитов Nb/Al и Nb/Si // Физика и техника высоких давлений, 2013, т. 23, № 1, с. 99–107.
4. Valeriy P. Korzhov, Michael I. Karpov, Dmitriy V. Prokhorov. Structure evolution of multilayer materials of heat-resistant intermetallic compounds under the influence of temperature in the process of diffusion welding under pressure and their mechanical properties // J. Internat. Scientific Public.: Materials, Methods and Technologies, 2013, v. 7, part 1, p. 343–361.
5. Karpov Mikhail I., Korzhov Valery P., Prokhorov Dmitry V., Zhelytyakova Irina S., Stroganova Tatiana S., Vnukov Victor I. Preparation, structure and high temperature properties of layered Nb/Al- and Ti/Al-composites // J. Intern. Scient. Public.: Materials, Methods and Technologies, 2014, v. 8, p. 177–185.
6. Karpov M.I., Vnukov V.I., Korzhov V.P., Stroganova T.S., Zhelytyakova I.S., Prokhorov D.V., Gnesin I.B., Kiiko V.M., Kolobov Y.R., Golosov E.V., Nekrasov A.N. Structure and mechanical properties of an eutectic high-temperature Nb–Si alloy grown by directional solidification // Russian metallurgy (Metally). 2014, t. 14, № 4, с. 267–274.
7. Стrogанова Т.С., Карпов М.И., Коржов В.П., Внук В.И., Прохоров Д.В., Желтякова И.С., Гнесин И.Б., Светлов И.Л. // Влияние титана и молибдена на структуру и механические свойства in-situ композита на основе системы ниобий-кремний. Известия РАН. Серия физическая, т. 80, №9, 2015, с. 1302–1306.
8. Светлов И.Л., Кузьмина Н.А., Нейман А.В., Исходжанова И.В., Карпов М.И., Стrogанова Т.С., Коржов В.П., Внук В.И. // Влияние скорости кристаллизации на микроструктуру, фазовый состав и прочность in-situ композита Nb/Nb₅Si₃. Известия РАН. Серия физическая, т. 80, №9, 2015, с. 1296–1301.
9. Jain P., Kumar K.S. Tensile Creep of Mo-Si-B Alloys // Acta Mater., 2010, v. 58, N 6, p. 2124–2142.
10. Брок Д. Основы механики разрушения / – М.: Высшая школа, 1980. – 368 с.