



DOUBLE-LAYER SURFACING COMPOSITIONS BASED ON FILLING MATERIAL OF Cr–Ti–C ALLOYING SYSTEM

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It was proposed in the work to use the method of furnace surfacing to produce double-layer macroheterogeneous compositions. It consists in successive impregnation of hard filler-alloy particles by two metal binder-alloys having different melting points. As a filler the Cr–Ti–C alloy is used, as binders the dispersion-hardening alloy of grade MNMts20-20 and hypoeutectic Fe–B–C alloy are used. As a result of impregnation a double-layer composition is produced, the layers of which are hardened by particles of the same filler, but having different composition of binder-alloy. It is shown that after furnace impregnation almost defect-free structure of double-layer compositions is formed, the porosity of which does not exceed 5–7 %. The comparison of dissolution rate of primary and peritectic phase of filler-alloy in melts based on copper and iron was made. Determined were the mechanical properties of layers of surfacing composition intended for hardening the parts operating under the conditions of non-uniform distribution of temperatures, stresses and deformations. 5 Ref. 3 Tables, 6 Figures.

Keywords: *furnace surfacing, double-layer surfacing compositions, interphase interaction, dissolution rate, mechanical properties*

The surfacing composite materials are rather widely applied for hardening the surface of parts operating under the conditions of intensive abrasive wear. Among the existing methods for producing composite deposited layers the best known is the method of furnace impregnation developed by the specialists of the E.O. Paton Electric Welding Institute [1]. This technology is distinguished by simplicity, sufficiently high efficiency and relatively low power consumption. In addition, it allows producing double- and multi-layer surfacing compositions [2].

The surfacing composite material is composed of particles of hard filler-alloy and metal binder-alloy. Moreover, the binder-alloy should produce a good wetting of particles of hard filler-alloy and surface of the part being surfaced. Due to macroheterogeneous structure the deposited layer acquires new operational properties, which are not inherent to each of its separate initial materials. As fillers in composite surfacing alloys the carbides, borides, nitrides and also crushed alloys of VK and TK type are used [3]. A challenging filler of composite materials is alloy 70 % Cr–20 % Ti–10 % C with high mechanical properties [4], which in case of right selection of the binder-alloy should provide a strong bonding with filler.

Therefore, the aim of this work was the study of structure of interfaces between the structural

components of double-layer macroheterogeneous compositions, mechanical properties of deposited composite layers produced by furnace surfacing, where as filler the alloy 70 % Cr–20 % Ti–10 % C is used, and as binder-alloys the manganese nickel silver of grade MNMts 20-20 and alloy Fe–3.1 % B–0.05 % C were used.

Experiment procedure. For surfacing of double-layer composition into the form repeating surface of deposited part with a gap equal to thickness of deposited layer, filler granules 70Cr–20Ti–10C of 0.2–2.5 mm were poured and compacted. Atop the granules the binder-alloys were placed separated by a partition.

Heating of form in the furnace during impregnation was carried out in two stages. At first it was heated to the temperature by 50 °C higher than melting point of more easily melted binder-alloy MNMts 20-20 and soaked for 15 min. Then, temperature in the furnace was increased by 50 °C higher than melting point of more refractory binder-alloy Fe–B–C and again soaked at this temperature for 15 min. At the end of impregnation the form was cooled together with the furnace. After removal of the form a double-layer composition was produced, the layers of which were hardened by the same filler but with different composition of binder-alloy.

The microstructure of deposited layers was studied in optical microscopes Neophot, GX-51 and Epytype-2. The quantitative metallographic examinations were carried out in diffraction analyzer Epiquant. The identification of phase components was performed using X-ray diffraction

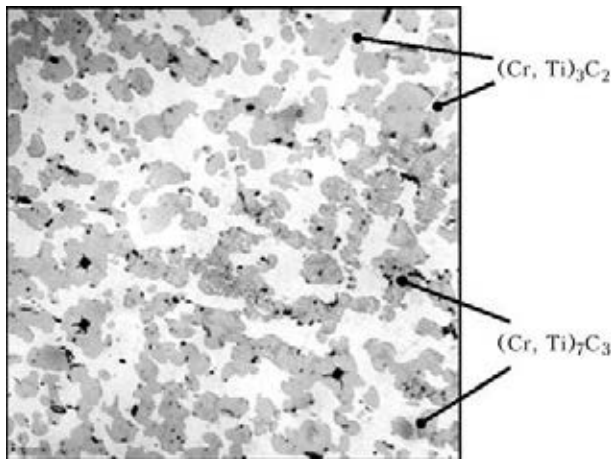


Figure 1. Microstructure ($\times 300$) of carbide Cr-Ti-C

analysis by filming the powder diffraction patterns in devices DRON-UM1 and HZG-4A in the filtered K_{α} -iron and copper radiation. The examinations using method of X-ray spectral microanalysis (XSMA) were performed in devices MS-46 with four channels of recording the intensity of radiating elements and in scanning microscopes-microanalyzers REMA 102-02M and Camebax. As an adapter device the latter was equipped with semiconductor detector LINK 860, series II. The linear rate of phases dissolution of filler in the molten metal was evaluated as the ratio of width of contact interaction areas (3 kV), formed at the interfaces, to the impregnation duration.

The experiment results. Before impregnation, filler-alloy Cr-Ti-C has two-phase structure consisting of dark-colored carbide crystals $(Cr, Ti)_3C_2$ on the background of bright carbide crystals $(Cr, Ti)_7C_3$ (Figure 1). Before impregnation, binder-alloy Fe-3.1B-0.05C has hypoeutectic structure, where at room temperature the crystals α -Fe and eutectic Fe-Fe₂(B, C) are observed

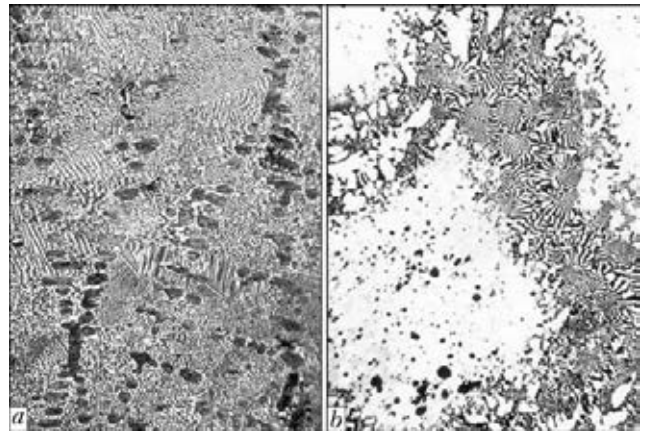


Figure 2. Microstructure ($\times 200$) of binder-alloy Fe-3.1 B-0.05 % C (a) and layer of composition (Cr-Ti-C)/(Fe-B-C) (b)

(Figure 2, a). In the structure of manganese nickel silver MNMTs 20-20 a solid solution of manganese and nickel on copper base is formed, in which the dispersed inclusions of intermetallic phase NiMn are present [5].

After furnace impregnation the structure of deposited layers is characterized by a uniform distribution of filler granules Cr-Ti-C (Figure 2, b). The average filler content is 60 ± 2 %. At the interface between layers the defects of lack of fusion type are absent. This is confirmed by the results of porosity determination in the compositions, which is not higher than 5-7 %.

In layer (Cr-Ti-C)/(Fe-B-C) at the interfaces between carbide Cr-Ti-C and Fe-based binder the zones of coarse inclusions (ZCI) occur characterized by the presence of carbide inclusions $(Cr, Ti)_3C_2$ and $(Cr, Ti)_7C_3$ in eutectic Fe-Fe₂(C, B) alloyed with chromium and titanium (Figure 3, Table 1). According to the results of XSMA after wetting with melt Fe-3.1B-0.05C the iron content in the surface layer of

Table 1. Structure and width of ZCI formed between the filler Cr-Ti-C and binders based on copper and iron

Binder	Initial structure of binder	Structure of interfaces on the side of		Width of ZCI, μm
		matrix	filler	
Fe-3.1 % B-0.05 % C	α -Fe, eutectic Fe-Fe ₂ (B, C)	α -Fe, Fe-(Fe, Cr, Ti) ₂ (B, C)	$(Cr, Ti)_3C_2$, $(Cr, Ti, Fe)_7C_3$, Fe-(Fe, Cr, Ti) ₃ (C, B)	75-80
MNMTs 20-20	Solid solution (Cu, Ni, Mn)	Solid solution (Cu, Ni, Mn, Cr, Ti, C)	Solid solution of Ni, Mn, Cu in $(Cr, Ti)_7C_3$, $(Cr, Ti)_3C_2$	7-10

Table 2. Results of determination of microhardness (GPa) of structural components of deposited layers with filler Cr-Ti-C and binders based on copper and iron before and after impregnation

Binder	Matrix		Filler				
	Initial alloy	Interface	Initial alloy		Interface		
			$(Cr, Ti)_7C_3$	$(Cr, Ti)_3C_2$	$(Cr, Ti)_7C_3$	$(Cr, Ti)_3C_2$	Fe-Fe ₂ (C, B)
Fe-3.1 % B-0.05 % C	4.7 \pm 0.1	5.8 \pm 0.1	19.5 \pm 0.5	25.1 \pm 0.6	17.8 \pm 0.4	25.0 \pm 0.5	8.1 \pm 0.3
MNMTs 20-20	0.73 \pm 0.4	0.97 \pm 0.2	19.5 \pm 0.5	25.1 \pm 0.6	18.9 \pm 0.3	24.7 \pm 0.7	-

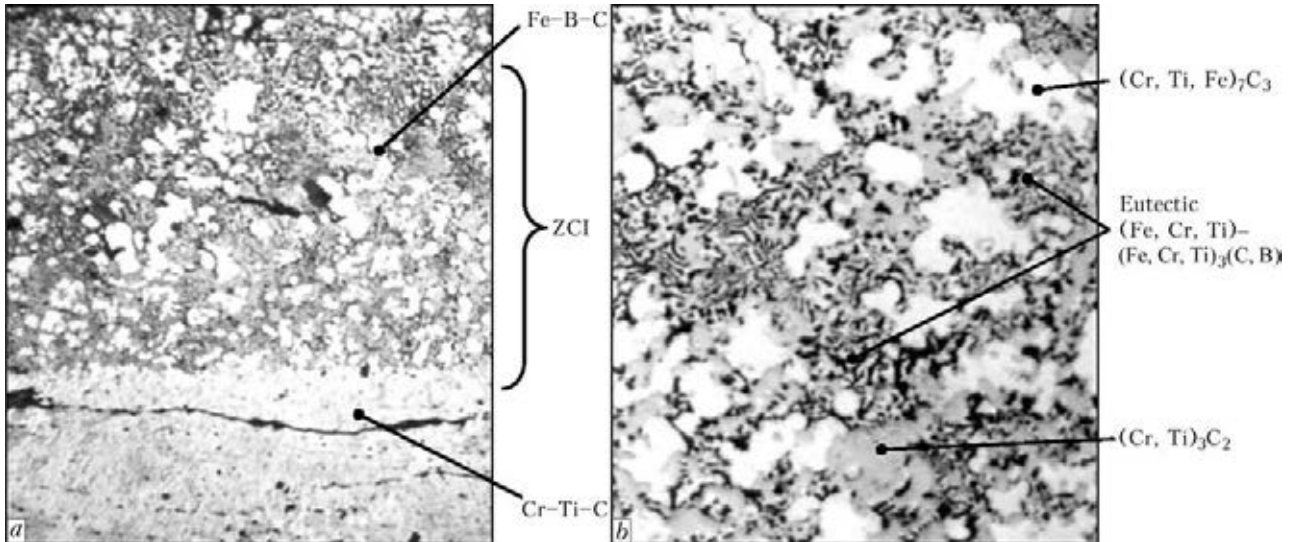


Figure 3. Microstructure of ZCI between the carbide Cr-Ti-C and binder-alloy Fe-3.1 % B-0.05 % C: *a* – $\times 300$; *b* – $\times 1000$

carbides $(\text{Cr, Ti})_7\text{C}_3$ increases up to 1.3 % (Figure 4). The composition of carbides $(\text{Cr, Ti})_3\text{C}_2$ corresponds to the initial one.

In the crystallized matrix in the vicinity of interface 0.2 % Cr and about 0.1 % Ti are additionally present. The concentration of chromium in carbides $(\text{Cr, Ti})_3\text{C}_2$ and especially in carbides $(\text{Cr, Ti})_7\text{C}_3$, observed in the structure of ZCI on the side of filler, is decreased, while that of iron is increased (see Figure 4). During moving the probe towards the filler the content of chromium in carbide phases is gradually increased whereas concentration of iron is decreased. At removal from ZCI towards the matrix the iron content is increased. The matrix of composite materials has structure of eutectic Fe-Fe₃(C, B) alloyed with chromium and titanium, which differs from the

initial structure of binder, where α -Fe and eutectic Fe₂(B, C) were present. Microhardness of matrix of composite materials is 4.7–5.8 GPa depending on the content of alloying elements (Table 2). In the vicinity of interface the microhardness of carbides $(\text{Cr, Ti})_7\text{C}_3$, additionally alloyed with iron, is reduced as compared to this characteristic of carbides of initial alloy Cr-Ti-C. The microhardness of carbides $(\text{Cr, Ti})_3\text{C}_2$ is not almost changed.

At the contact interaction of Cr-Ti carbide with the melt based on iron the processes of strong chemical interaction are observed at the interface, characterized by carbides $(\text{Cr, Ti})_7\text{C}_3$ dissolution in the molten metal. The molten metal penetrates mainly along the grain boundaries of carbides $(\text{Cr, Ti})_7\text{C}_3$, the diffusion of iron at-

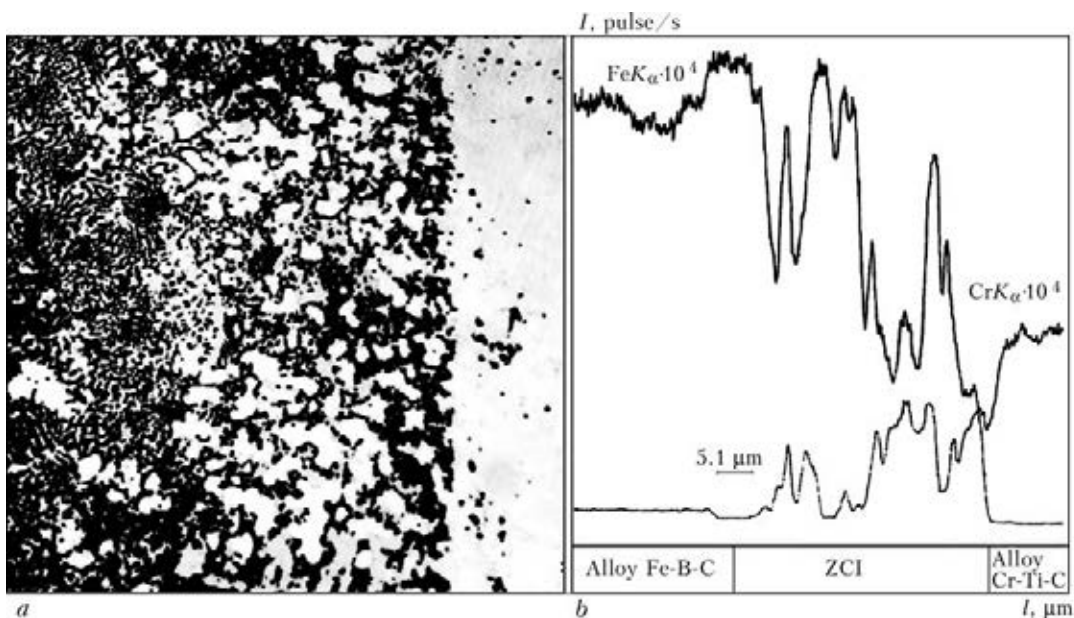


Figure 4. Microstructure (*a* – $\times 300$) and diagram of distribution of radiation intensity of elements during moving the probe across the area of boundary interface (Cr-Ti-C)/(Fe-B-C) (*b*)



Table 3. Results of determination of mechanical properties of deposited layers

Layer	Hardness HRC	Total micro-hardness, GPa	Compression strength, MPa
(Cr-Ti-C)/MNMts 20-20	43±3	8.1±0.3	3105±5
(Cr-Ti-C)/(Fe-B-C)	60±2	10.8±0.4	2240±10

oms along which is facilitated. Dissolution of carbide $(Cr, Ti)_7C_3$ alloyed with iron occurs at rate of 2.0–2.5 $\mu\text{m}/\text{min}$ in a short period of time. Rapid dissolution of carbides $(Cr, Ti)_7C_3$ may be also facilitated by transformation of carbide $(Cr, Ti)_7C_3$ into lower carbide $(Cr, Ti)_{23}C_6$, caused by loss of carbon in diffusion.

Carbide $(Cr, Ti)_3C_2$ is dissolved in the molten matrix at rate of 0.4–0.7 $\mu\text{m}/\text{min}$. Moreover, penetration of iron atoms from the melt to this carbides is almost absent. Carbides $(Cr, Ti)_3C_2$ are preserved almost unchanged during impregnation and are present in the form of dark inclusions in eutectic Fe-Fe₃(C, B) alloyed with chromium and titanium.

In the structure of layer (Cr-Ti-C)/MNMts 20-20 at the interfaces with matrix the surface of hard alloy particles does not change (Figure 5). However, XSPA and microdurometric measurements confirm the formation of ZCI between the filler and matrix (see Tables 1 and 2). Titanium and chromium from Cr-Ti carbide penetrate into nickel silver melt (Figure 6). Nickel and manganese diffuse from the molten metal into the hard alloy mainly along the grain boundaries. In presence of these elements the diffusion of copper into carbide is also facilitated. At the interfaces the processes of weak chemical interactions occur, which include the stage of Cr-Ti carbide dissolution in nickel silver at a rate of 0.2–0.6 $\mu\text{m}/\text{min}$.

At the interface between the layers of surfacing composition on the side of binder MNMts 20-20 crystals of Fe-based phase alloyed with copper and nickel appear additionally. The structure of eutectic binder is almost unchanged. In the transition from the layer with alloy-binder MNMts 20-20 to the layer with binder-alloy Fe-B-C the general microhardness is 1.3 times increased, hardness is increased by 1.4 times, and the compression strength is reduced by 1.4 times (Table 3).

Discussion of the results. The sequence of processes occurring during impregnation of filler Cr-Ti-C with melts on copper and iron base in a generalized way can be presented in the following way. At first the molten binder penetrates



Figure 5. Microstructure ($\times 200$) of layer of composition (Cr-Ti-C)/MNMts 20-20

into the gaps between the freely poured filler particles, fills the pores, cracks and other defects. Then, mutual diffusion of binder and filler components occurs. In case of wetting Cr-Ti carbide with molten nickel silver, penetration of filler components into the binder is mainly observed.

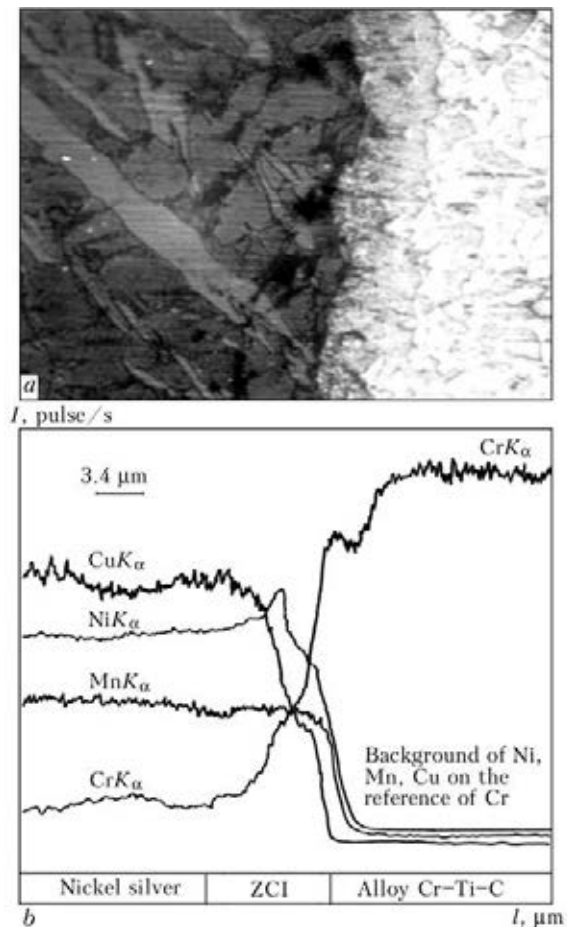


Figure 6. Microstructure ($\times 200$) (a), and diagram of distribution of radiation intensity of elements during moving the probe across the area of interface (Cr-Ti-C)/MNMts 20-20 (b)



The filler phases are dissolved in nickel silver at the slowest rate. Therefore, after solidification of composite materials with Cu-based matrix at the interfaces the changes in structure are absent, and only using XSM method it is possible to detect the presence of filler elements in the matrix, and vice versa.

The limit of mutual saturation by binder and filler components during wetting with Fe-based melts is not achieved during the impregnation process. In this case, the binder components penetrate into the hard alloy in a much larger quantity. The solubility of iron in phase $(Cr, Ti)_7C_3$ of Cr-Ti carbide is 10 times higher than that of nickel and manganese, which diffuse from the nickel silver binder. Moreover, as compared to nickel silver, the molten iron binder penetrates along the grain boundaries of the carbide $(Cr, Ti)_7C_3$ to a greater depth and almost completely dissolves carbides $(Cr, Ti)_7C_3$, which are present in the surface layer of filler. Carbides $(Cr, Ti)_3C_2$ are dissolved in the molten metal at a much slower rate.

In the liquid state at the surfaces between the filler and matrix the diffusion processes occur. During subsequent cooling due to the limited solubility of components, carbide crystals $(Cr, Ti)_7C_3$ alloyed with iron are precipitated from the melt. Their origination is facilitated on the surface of non-dissolved phase $(Cr, Ti)_7C_3$. The remaining liquid is crystallized with formation of eutectic Fe-Fe₃(C, B) alloyed with chromium and titanium, along the grain boundaries as a result of enrichment of the melt, corresponding to composition of initial eutectic Fe-Fe₂(B, C), with carbon, chromium and titanium during dissolution of Cr-Ti carbide.

At the interfaces between the layers of double-layer composite coating also dissolution and diffusion processes occur. New phases appear only in the transition zone between the layers on copper and iron base. They represent Fe-based solid solution formed during cooling due to the limited solubility of components in copper.

In conclusion it may be noted that the method of furnace impregnation can be used for surfacing of double-layer compositions hardened by filler of the same composition but having different composition of matrix. The investigations of physical and chemical processes occurring during formation of interfaces between the filler and matrix in the structure of deposited layers show prospects in application of alloy Cr-Ti-C as a filler, and metal alloys MNMs 20-20 and Fe-3.1 % B-0.05 % C as a binders in double-layer composition. It is characterized by almost defect-free structure and presence of mechanical properties gradient along the cross section. It allows using the proposed composition for strengthening the surface of parts operating under non-uniform temperature distributions, stresses, deformations, etc.

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