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INVESTIGATION OF MACHINING PROCESSES

Technological Principles of Plasma Nanostructuring of Tool Composites for Instruments with Complex Surface Shapes

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Abstract—Based on the results of studying the structure and properties of tool steels (R6M5 and Kh12M) and hard alloy VK8 used in threading tools (cutters, combs, and rollers), it is determined that a nanocrystalline structure can be formed in the modified region formed during surface treatment with a highly concentrated plasma jet. The formation of nanodisperse particles of martensite (in steels) and carbides (in steels and hard alloys) is caused by an increased cooling rate and manifestation of the barrier effect of dispersion hardening under conditions of rapid crystallization (microfusion treatment) or rapid quenching (nonmelting treatment). The effect of plasma treatment on the cutting surface of grinding wheels made of superhard materials with metallic binders is studied. The change in the elemental surface composition, the presence of an oxygen-containing coating film on the binder and superhard material grains, and the change of the hardness of the binder surface are shown.

Keywords: plasma surface modification, tool steel, hard alloy, wheels based on superhard materials, surface with complex shape

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Among the methods used for the formation of directed dispersion in the structure of various materials, the treatment of the surface with highly concentrated heat sources (HCHS), such as the laser beam and plasma jet [1, 2], is a promising one. To date, considerable experience has been gained in the development and application of the laser and plasma surface nanostructuring of steels and alloys [3–8]. Among the treatment methods, the plasma processing of materials with HCHS has certain technical and economic advantages [9].

The general principles of formation of a fine-grained structure in steels and alloys during the HCHS treatment consist in achieving very high heating and cooling rates (10^5–10^6 °C/s) when moving to temperatures exceeding the critical temperature of phase transformations (for steels, A_{c3}). This condition is necessary, but not sufficient for the formation of surface nanolayers in steels [8]. According to experience gained from [3–8], the cooling rate during the HCHS processing should be one to two orders of magnitude higher than that used during surface hardening in order to obtain a nanocrystalline structure (the mean size of structural parameters, i.e., martensite and carbide particles, is about 100 nm [10, 11]). In general, the application of a certain technological version of plasma modification depends on the ratio between the main adjustable thermophysical parameters of plasma heating, which are maximum heating temperature T of the surface layer and cooling rate W. Figure 1 shows the regions of application of plasma technologies.

At present, considerable experience has been gained in the development of optimal technologies for plasma hardening of metalworking tools (see Fig. 1, region *2*) in accordance with their shape and size, and with the composition of the tool material [9, 12, 13]. However, the application of plasma nanostructuring of the tool (region *3*) requires either the modernization of plasma torches [14] or the development of new technological approaches.

As was established in [15], the highest hardness and heat resistance values during plasma modification of high-performance steels are achieved in the case of plasma modification with surface micromelting (Fig. 2) when the heating temperature in a thin surface layer (up to 0.1 mm) exceeds the melting temperature. In this case, the structure of the micromelted region consists of ultrafine lamellar martensite, highly

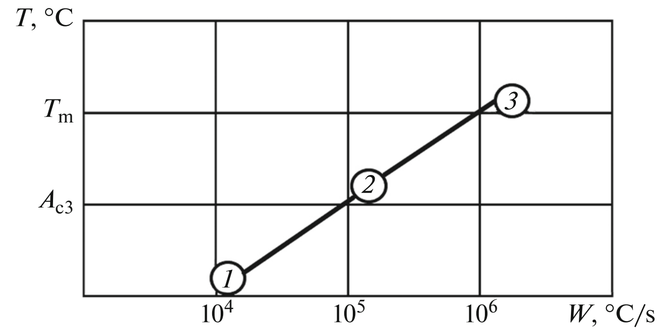

Fig. 1. Regions of application of plasma modification technology versions: (*1*) plasma tempering; (*2*) plasma hardening; (*3*) plasma nanostructuring.

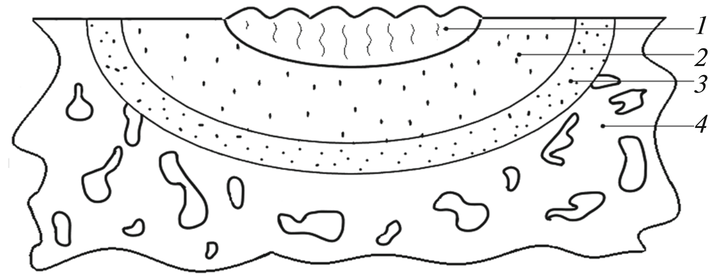

Fig. 2. Scheme of the structure of the region modified by plasma treatment with surface melting: (*1*) melted zone; (*2*) hardened zone; (*3*) transient zone (zone of incomplete hardening); (*4*) initial structure.

supersaturated residual austenite (46%), and dispersed eutectic carbide particles with a globular shape that do not form a solid framework. Diffusion mass transfer (mainly carbon) on the austenite growth front does not have time to complete, which limits the formation of the γ phase. This leads to the degeneration of dendritic crystallization forms and the layer of plasma micromelting has a cellular structure. As with other methods of rapid crystallization [16], each grain has a block structure and the block size is about 100 nm.

As has been considered until recently [15, 17, 18], the effect of surface nanostructuring during the HCHS treatment can be obtained only under conditions that provide surface micromelting due to qualitative changes in the crystallization mechanism (appearance of cellular structures instead of dendritic ones) and the freezing of new metastable phases. However, the possibility of laser nanostructuring of the surface without melting the material was established in [3]. This is possible in the case of surface treatment of complex teeth-shaped structures. A common example of such surfaces is a thread-forming tool, such as combs, rollers, cutters, taps and dies, etc. For example, the working part of the threading comb has six to eight steps of threading, and one and half to two of them are located on cutting part l_1 and four and half to six of them are steps located on calibration part l_2 (Fig. 3a).

The working surface of the grinding tool (Fig. 4) made of superhard materials (SHMs) also has the same complex tooth-shape surface.

To determine the possibility of surface nanostructuring during heating of complex surfaces of metal-working tools, thread cutters with plates made of hard alloy VK8, threaded combs made of steel R6M5, threaded rollers made of steel Kh12M, and cutting surfaces of grinding wheels made of SHM were processed with a highly concentrated plasma jet. The aim of this study was to identify common features of plasma modification of such complex surfaces.

Plasma treatment regimes for different technological versions were chosen according to the published recommendations [9, 12]. Measurements of the values of hardness *HV*, determination of the heat resistance of phase compositions and the crystal lattice parameters, and metallographic studies were performed according to standard methods on instrument samples. Parameters *T* and *W* in the plasma modification process are complex parameters, rather than directly controlled parameters. On the one hand, these are the parameters of the plasma treatment regime (current *I*, voltage *U*, pressure *p*, plasma-forming

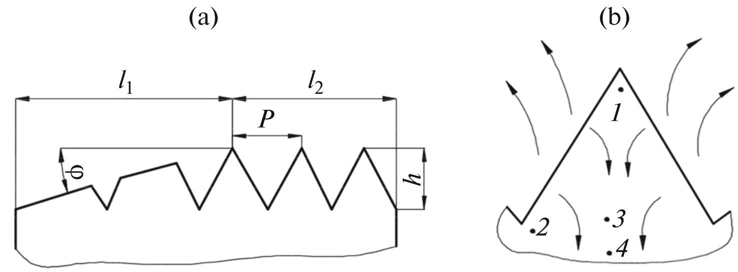

Fig. 3. (a) Working part of the threaded comb and (b) the scheme of heat flow movement directions during rapid cooling in the course of plasma processing; points *1–4* are the points of metallographic and X-ray structural studies.

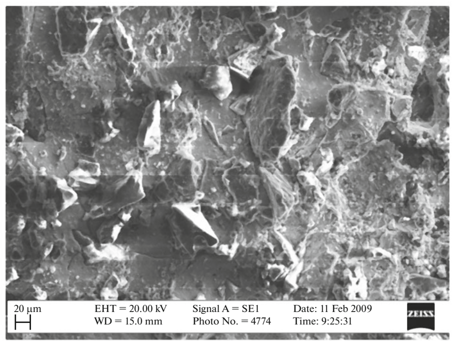

Fig. 4. Cutting surface of a diamond wheel with protrusions of diamond grains and depressions between them.

gas flow rate Q, etc.); on the other hand, they are the parameters of the shape and size of the tool subjected to surface hardening.

The plasma treatment of thread-cutting tools made of high-performance steels both can be applied to the finished tool that has passed standard bulk heat treatment (hardening and tempering) at the stage of manufacturing and can be introduced into the technological process of heat treatment instead of bulk quenching and subsequent tempering or in combinaion with them.

The complex rear surface of prismatic combs made of steel R6M5 (see Fig. 3a) for cutting a metric thread with a step of $P = 4$ mm, which had passed standard bulk heat treatment at the stage of manufacturing, i.e., the quenching from 1220°C and three-fold tempering at a temperature of 560°C (exposure to this temperature for 1 h), was subjected to plasma treatment. The following two versions of complex technology have been studied: (A) plasma modification after standard bulk heat treatment and (B) the same modification with subsequent three-fold standard bulk tempering. Templates were cut from the treated combs, and specimens were prepared for metallographic and X-ray diffraction structural studies at points *1–4* (see Fig. 3b). The study results are given in Table 1.

The highest values of heating temperature (close to, but not exceeding the melting temperature) and the highest cooling rates are achieved in the bulk of metal at the tips of the teeth (see Fig. 3b, point *1*). Slightly lower values of these parameters are achieved in the bulk of metal in the depressions (point *2*), which were further reduced in the case of moving the modified zone deep into the bulk (point *3*) to the metal with the initial structure (point *4*). The highest degree of structure dispersion of the modified region and the highest hardness values are achieved at the tips of the teeth and in the depressions. In the case of moving deep into the modified zone, the dispersion of the structure and the hardness are slightly reduced, but remain much higher than in the initial metal.

If the combs are strengthened according to version B (with the final bulk tempering), then the properties are substantially improved (see Table 1). This is due to the fact that two processes—namely, the decay of residual austenite (increasing the content of the martensitic phase) and the dispersion hardening of the modified region of metal (increasing the content of the carbide phase)—occur simultaneously in the case of tempering modified steel R6M5 (see Table 1). The released secondary carbides have an extremely high degree of dispersion, especially at the edges (Figs. 5a, 5b), and are evenly distributed in the martensitic matrix. A high density of dislocations is also an important factor during the dispersion hardening of the

Table 1. Hardness and properties of the metal of the modified region on the threaded combs made of steel R6M5

Treatment version	Studied point	Hardness *HV*	Heat resistance, °C	Phase composition, %			Crystal lattice parameters		
				martensite	austenite	carbides	martensite lattice period *a*, nm	dislocationden sity $\rho \times 10^{-16}$, m^{-2}	block size $D \times 10^7$, m
Initial state	4	820–840	620	86	2	12	0.2868	4.5	1.112
A	1	905–920	645	55	42	3	0.2875	37.2	0.153
	2	910–915	655	60	35	5	0.2874	35.7	0.160
	3	900–915	645	88	2	10	0.2874	35.2	0.164
B	1	1000–1025	690	86	2	12	0.2873	45.0	0.091
	2	990–1015	690	86	2	12	0.2872	43.4	0.101
	3	970–990	680	88	2	10	0.2872	42.1	0.107

modified region of metal. The period of the martensite crystal lattice decreases and the block size increases after tempering (see Table 1), but the mechanisms of dislocation and, especially, dispersion strengthening operate more effectively. The hardness and heat resistance values exceeding the levels attained in a massive steel tool by plasma modification without melting the surface [9] and corresponding to the levels for plasma treatment with surface micromelting were achieved.

Threaded rollers have a similar complex surface shape (see Fig. 3). Since the curvature radius or the size of the flat section of their vertices is very small [19], the model shown in Fig. 3b was adopted as a theoretical thermal model of plasma heating of such a surface. High-chromium Kh12M steels similar in nature to high-performance steels are used to make such cold-deforming tools, since they undergo the same transformations as high-performance steels [20], but the carbide phase in the Kh12M steel substantially differs in composition and properties, which has an effect on the choice of optimal regimes for the bulk and surface heat treatment.

The effect of plasma modification without surface melting (in order to preserve geometrical parameters) on thread rollers made of steel Kh12M has been studied. Modification versions without tempering and with tempering at a temperature of 200°C are considered. The microstructures at points *1* and *4* (by analogy with Fig. 3b) are shown in Fig. 6. In metallographic studies with high magnification at the points tangent to the tooth vertices (point *1*, see Figs. 3b and 6a), the formation of a nanocrystalline structure with average sizes of the particles of carbides and lamellar twin martensite in the range of 50–150 nm (see Fig. 6a) is clearly observed (formed as a result of the decay of ultrafine austenite in the process of rapid quenching). During subsequent tempering, the metal of the modified region in these volumes is further strengthened by dispersion hardening caused by the release of secondary nanoparticles of carbides (see Fig. 6b). The hardness at point *1* is *HV* 945–965 after plasma modification and *HV* 995–1015 after subsequent tempering, which almost correspond to the values achieved for steel R6M5 by plasma nanostructuring.

The cooling after plasma heating at point *3*, i.e., at a distance from the tooth tip, is mainly caused by heat dissipation to the material of the roller. The structure at this point (see Figs. 6c and 6d) and the hardness values (*HV* 840–855; after tempering, *HV* 870–915) are similar to those of massive samples made of steel Kh12M after plasma modification without melting [9, 12].

Thus, we can conclude from the above that the hardness (and heat resistance for steel R6M5) of tools made of high-performance or high-chromium steels may increase under certain plasma modification versions because of the formation of a nanocrystalline martensitic carbide structure in the modified region after finishing tempering. This can be the processing of a massive tool made of high-performance steel by the micromelting method, in which a nanocrystalline structure of the matrix is formed in the process of ultrafast crystallization of the molten layer. The plasma treatment of the surface with a complex (toothed) shape (see Fig. 3b) is another case, in which a nanocrystalline structure of the matrix in the solid phase is formed in the process of rapid quenching (10^6–10^7°C/s). The dispersed carbide particles released in both cases are extremely small and act as barriers to the growth of austenitic grains and, consequently, martensite crystals during rapid quenching from both liquid and solid phases.

In addition to the strengthening of steel tools, there is information about the prospects of plasma processing to improve the performance of carbide tools [9, 12, 21]. As was established previously [12, 22], the

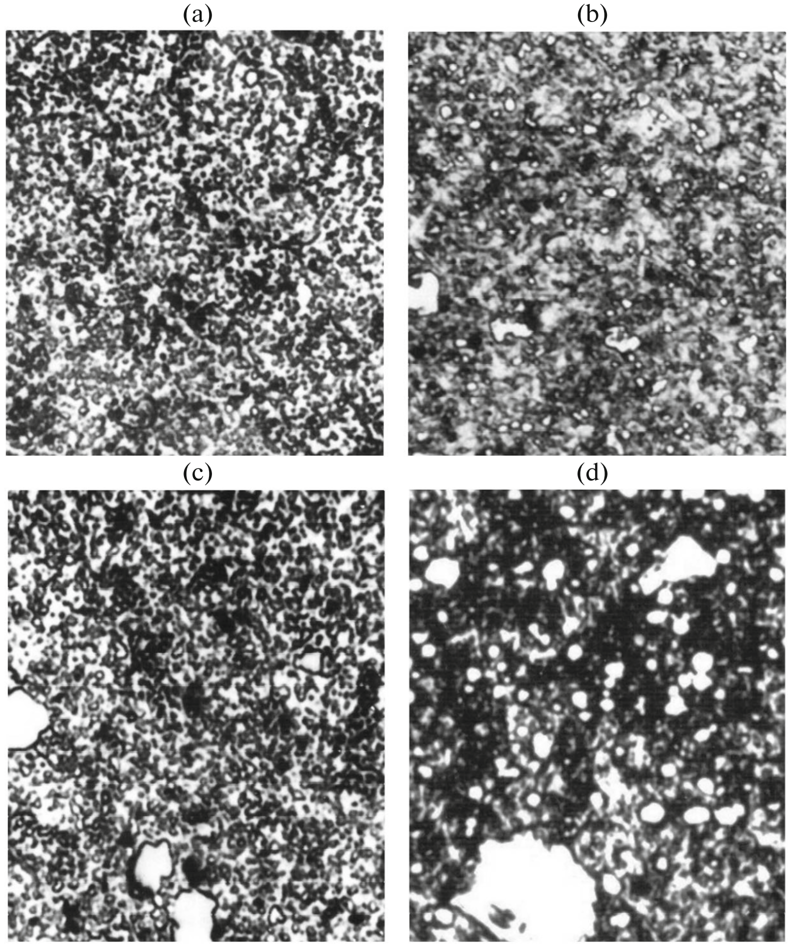

Fig. 5. Microstructure of the metal of the modified region on the combs made of steel R6M5 after hardening according to heat processing version B at the points. Figure parts (a–d) correspond to points *1–4* in Fig. 3b; magnification ×1000.

highest hardness values in the studied alloys are achieved by plasma treatment without melting the binder, namely, by heating to temperatures of about 1300°C. In this case, the contact melting of the interfacial carbide–binder boundaries is a main process that determines the structure and properties of the reinforced zone. Conditions are created for obtaining hard alloys with an ultrafine composite structure and a high level of performance properties. The possibility of plasma surface nanostructuring of carbide tools are studied using complex surfaces of sharpened threading cutters equipped with soldered carbide inserts as an example, which are the most common type of threading tools. The heating of such cutters with concentrated plasma (Fig. 7) creates thermophysical conditions that are similar to the conditions shown in Fig. 3b. The cooling rate is maximum at points that are close as possible to the tip of the cutter (Fig. 8) [23]. A study of the microstructure of the modified region on cutters with plates made of VK8 alloy treated by the optimal technology of plasma modification [21] showed that it has an inhomogeneous structure (see Fig. 8).

During plasma modification, the structure of hard alloys further develops under nonequilibrium conditions of rapid cooling. The initial carbide grains are disintegrated because of contact melting at the grain boundary [21] and subsequent decomposition of the supersaturated solid solution of W and C in Co with release of ultrafine secondary carbides. The fused grains of primary carbides acquire an irregular shape and loose edges, and the grains of secondary carbides that stand out as a result of dispersion hardening mostly become round shaped. The heating temperature decreases at a distance from the edges of the plate to the depth of the modified region (see Fig. 8a, points *2* and *3*); therefore, the contact melting region of primary carbides also contracts and the grain size increases, gradually moving to the size of grains in the initial structure. Along with partially melted grains, unmelted grains with a faceted shape are also preserved (see Fig. 8).

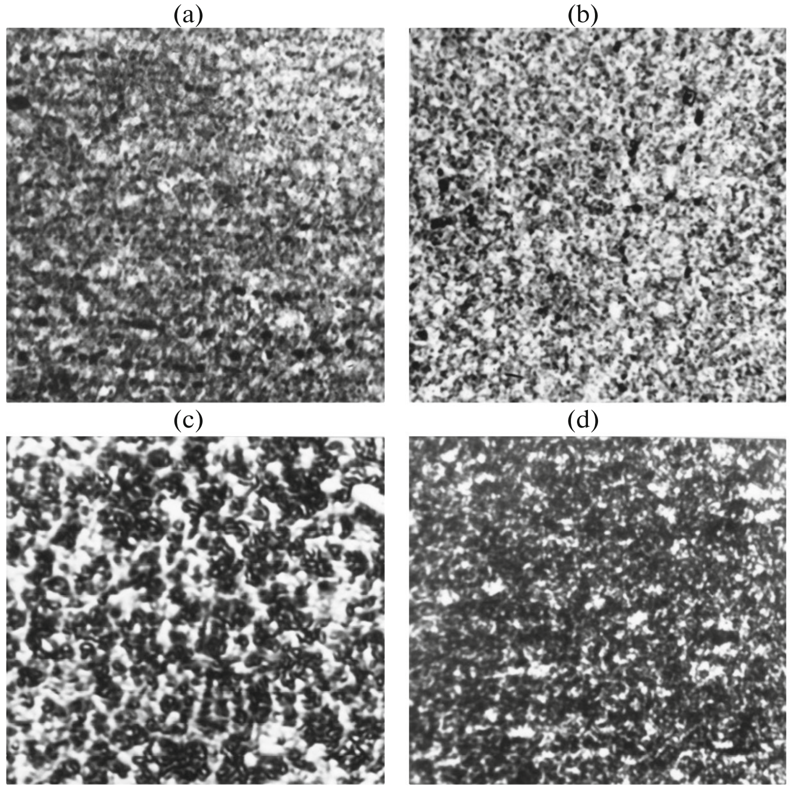

Fig. 6. Microstructure of individual points of the modified region on threaded rollers made of steel Kh12M after (a, c) plasma treatment and (b, d) subsequent tempering at a temperature of 200°C. Figure parts (a) and (b) correspond to point *1* and (c) and (d) to point *3* in Fig. 3b; magnification ×2000.

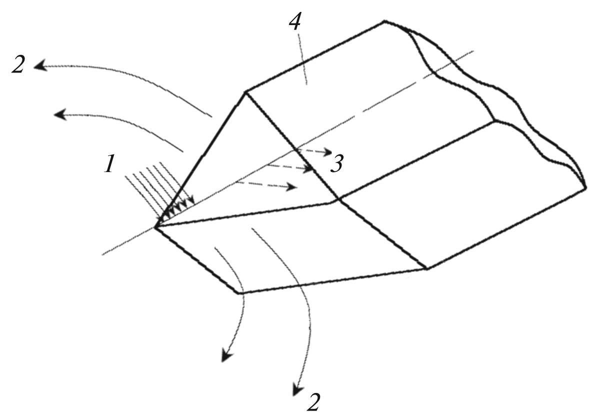

Fig. 7. Schematic representation of heat flow directions during heating of cutter *4* with plasma jet *1*; cooling flows in (*2*) the air and (*3*) cutter material.

Dispersion hardening of the binder supersaturated with alloying elements occurs upon rapid cooling with the release of ultrafine carbide particles. According to the results of X-ray diffraction analysis, not only the microstructure but also the substructure parameter is defragmented: size D of mosaic blocks decreases by a factor of more than 5 for alloys VK4 and VK8, and by a factor of almost 3.5 for alloy T5K10. In proportion to the decrease in the dispersity of the structure, the hardness of the metal of the modified region decreases to HV 1680, HV 1590, and HV 1400 at points *1–3*, respectively. Thus, metallographic and X-ray structural studies have determined the possibility of formation of a submicrocrystalline structure with an average carbide particle size of $d_c \approx 100$ nm and a high level of hardness in the modified region of hard alloys.

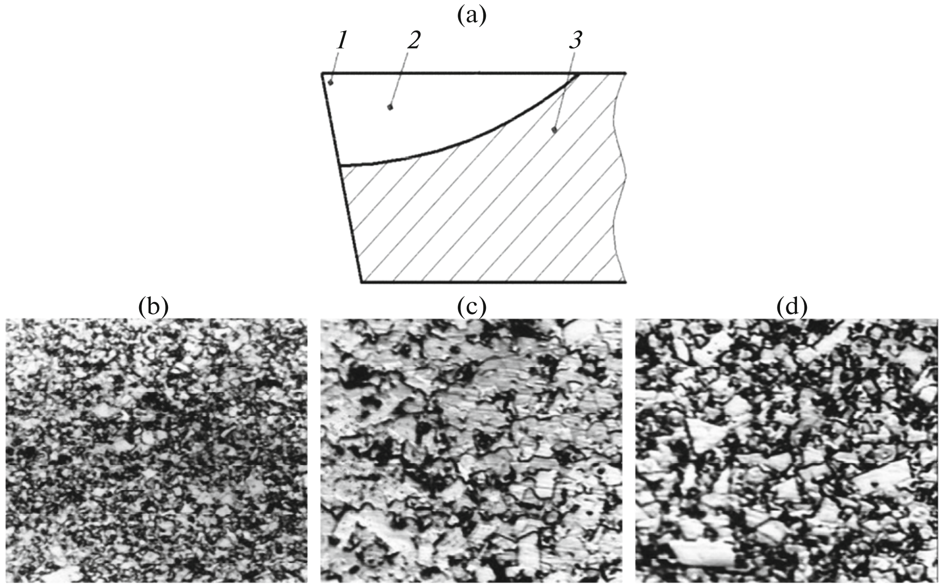

Fig. 8. (a) Layout of the cutter study points and (b–d) the microstructure of individual sections of the modified region on the threading cutters made of alloy VK8 at points *1–3*, respectively; ×2000.

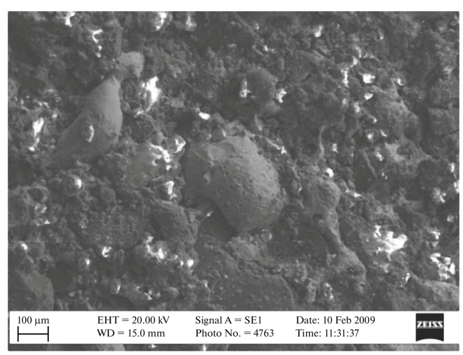

Fig. 9. The presence of regions with melting granules on the cutting surface of the wheel based on the MO20-2 metallic binder after intense exposure to plasma (Q = 18 kW, v = 25 cm/min).

It was investigated how this process occurs on the complex surface of diamond abrasive wheels (see Fig. 4). In the case of using metallic binders, similar changes in the surface structure occur in places of intense exposure to a plasma jet, which leads to melting of the binder (Fig. 9).

For such binders, the same as for polymer binders, the plasma flow rate should be low and the speed of movement of the plasma jet should be increased in order not to adversely affect the working surface of the wheel. If it is necessary to increase the effect of the plasma jet (for example, for binders such as M2-01), then the plasma flow rate should be increased and the speed of movement of the plasma jet should be reduced. Thus, both factors act as regulators of plasma intensity in metallic binders. This makes it possible to achieve the desired effect of opening the abrasive grains without cracking the binder. In this case, the metallic binder may form a coating on both SHM grains and the binder itself in the course of plasma processing (Fig. 10), which can change the contact processes in the processed region during grinding (distribution of elements is given in Table 2).

An analysis of the elemental composition of films that are formed on the abrasive grains and the binder in the region of plasma exposure (see Fig. 10) showed that they contain a lot of oxygen. Moreover, it is fixed to a larger extent on the grains (from 26.7 to 33.6 wt %) and to a lesser extent on the binder (from 16.9 to 25.0 wt %). The distribution of the binder elements is somewhat different in the films. If the Cu : Sn : Sb ratio between the main elements for the initial mixture of the MO20-2 binder is 8.75 : 2 : 1, then

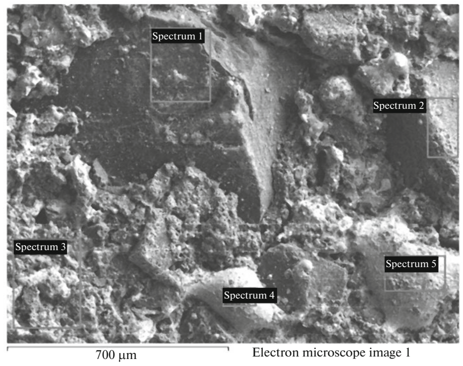

Fig. 10. Microphotograph of the surface of the grinding wheel based on the MO20-2 metallic binder with a coating after plasma modification.

Table 2. Distribution of elements on the working surface of the grinding wheel made of SHM with the MO20-2 metallic binder (see Fig. 4) after plasma modification

Spectrum	Elements, wt %								
	B	C	O	Na	Si	Fe	Cu	Sn	Sb
1	39.98	18.88	26.66	3.09	1.36	2.26	4.78	2.07	0.91
2	32.50	18.98	33.61	4.14	1.73	1.19	4.92	1.96	0.98
3	44.39	16.35	21.45	1.46	1.16	0.71	9.27	4.13	1.08
4	–	8.13	25.01	–	1.62	0.74	42.75	17.15	4.59
5	–	8.25	16.91	–	1.20	1.20	51.29	13.57	7.58

this ratio in the abrasive grains is 5 : 2 : 1 (reduced amount of copper). In turn, this ratio on the binder surface is 8.95 : 3.78 : 1 (the amount of tin in the film exceeds its amount in the initial mixture by a factor of 1.79).

Such redistribution of the elemental composition of the binder is revealed in [24] when studying the cross section of the binder in the well made by a plasma jet. Thus, the ratio between copper and tin is 4 in the depth of the binder and 2.88 to 1.28 on the surface of the well. This also indicates that the amount of tin in the surface layer of the binder increases after exposure to plasma. This redistribution is more characteristic of strong plasma exposure and also increases the hardness of the working layer of the wheel. Thus, the hardness of the working surface of the wheel based on the MO20-2 binder is 99 *HRB* in the absence of plasma exposure, the hardness after a strong plasma impact is 95–97 *HRB* at the edges of the well and 93 *HRB* in the center of the well.

CONCLUSIONS

In the narrow range of plasma surface modification parameters optimal for high-performance steels (micromelting), an ultrafine cast structure of martensite and carbides with an average particle size of about 100 nm is formed, which is caused by rapid quenching from the liquid state.

For the threading tool made of steels R6M5 and Kh12M with a complex tooth-shaped surface, direct plasma nanostructuring is possible in the case of plasma modification without melting because of rapid quenching in the solid state. In the regions adjacent to the tips of the cutting teeth, martensite and carbide nanoparticles with a size of 50–100 nm are formed. Carbide particles act as barriers to the growth of austenite grains and, consequently, martensite plates. During the next bulk tempering, the metal of the modified region is further strengthened by the release of secondary carbide nanoparticles.

With plasma modification of carbide cutters with a small sharpening angle, the cooling rate substantially increases in the places tangential to the tip of the cutter. The dispersion of the initial carbides (due to contact melting of their grain boundaries) and the release of ultrafine secondary carbides ($\bar{d}_c \approx 100$ nm) during the decomposition of the supersaturated solid solution of W and C in cobalt are observed.

To substantially improve the performance of the grinding tool, it is necessary to study the total effect of the impact on the binder, namely, the opening of the grains, the change of the elemental composition of the binder, the formation of antifriction films on the surface of the wheel, and the strengthening of the working layer of the wheel.

CONFLICT OF INTEREST

The authors declare that they have no conflicts of interest.

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