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SURFACING AT THE E.O. PATON ELECTRIC WELDING INSTITUTE

Department of Physical-Metallurgical Processes for Surfacing of Wear- and Heat-Resistant Steels of the E.O. Paton Electric Welding Institute is 50 years old. Organiser and the first manager (1954–1984) of this Department was Prof. Dr. (Eng.) I.I. Frumin, famous scientist in the field of metallurgy of welding and surfacing, founder of the surfacing school at the former USSR, laureate of the USSR State Prize and Evgeny Paton Prize.

The Department conducted the most extensive studies in the field of arc surfacing. The technology for wear-resistant surfacing of steel rolling mill rolls and flux-cored surfacing wire PP-3Kh2V8 (currently designated as PP-Np-35V9Kh3GSF), the first surfacing wire in the USSR, were developed early in the 1950s under the leadership of I.I. Frumin. Surfacing of rolling mill rolls using this wire, first practically applied at the Lenin Pipe Rolling Works in Dnepropetrovsk, yielded excellent results — life of the rolls increased approximately 10 times.

After such a success, the process of wear-resistant arc surfacing of steel rolls using flux-cored wire was introduced at the Magnitogorsk, Kuznetsk and Kommunarsk (now Alchevsk) Metallurgical Works, as well as at many other enterprises of the former USSR. Centralised production of the flux-cored wire was arranged at the Magnitogorsk Metalware-Metallurgical Factory, and commercial manufacture of the roll-surfacing machine tools was started at the Kramatorsk Machine-Tool Industrial Association.

The Department continued working in this area. To date it has developed a wide range of flux-cored wires for surfacing of different types of steel rolling mill rolls. This allows a customer to choose an optimal composition of deposited metal for each type of the rolls, based on their service conditions, character and intensity of wear. For about 50 years the flux-cored wire PP-Np-35V9Kh3GSF was and still is one of the most extensively used wires for surfacing of not only rolling mill rolls, but also other parts operating under similar conditions.

In addition to wires for surfacing rolling mill rolls, the Department has developed more than two tens of different grades of flux-cored wire, as well as technologies for submerged-arc, open-arc and gas-shielded surfacing of different machine parts and mechanisms operating under conditions of almost all known types of wear. Among the available flux-cored wires the most extensively used ones are PP-AN194 and PP-AN198 (for hard-facing of shafts, axles, crane wheels, etc.); PP-AN130, PP-AN132, PP-AN140, PP-AN148 (for hard-facing of cold and hot stamping dies); PP-AN158 (for hard-facing of hydraulic press rams); PP-AN159 and PP-AN174 (for hard-facing of continuous casting machine rollers); PP-AN192 and PP-AN197 (for hard-facing of parts operating under intensive abrasive wear conditions); and PP-AN105 (for hard-facing of steel 110G13L parts).

Much success has been achieved in the field of plasma surfacing. The Department developed and introduced into commercial application the plasma method for cladding discharge valves of internal combustion engines using additives in the form of vacuum-sintered rings of powders of nickel, chromium, graphite and other materials. A ring is placed into a groove of the valve plate and melted by the plasmatron. As a result, a layer of a heat-resistant alloy is formed on the working face of the valve.

Plasma cladding using powder additives (plasma-powder cladding) is the most flexible and versatile process, and the Department continues successfully upgrading this cladding method, retaining its leading position in the field. The Department developed powder additives on the iron, nickel, cobalt and copper base, technologies for cladding parts of general, power and petrochemical engineering equipment, screws of extruders for pressing plastics and rubber mixes, valves of internal combustion engines, knives for hot and cold cutting of metals, multi-blade cutting tools and other components, as well as specialised and versatile cladding machines.

The most important advantage of plasma-powder cladding is the possibility of producing minimal penetration: the share of base metal in the deposited one is no more than 5–10 %, which is especially important for cladding expensive nickel- and cobalt-base alloys.

The Department completed a number of interesting studies on electroslag surfacing (ESS). This process was used to solve the problem of surfacing rolling mill rolls of cast iron and hypereutectoid steels. The sectional current-conducting mould was developed for commercial implementation of ESS of cast iron rolls. The mould allows non-compacted materials of almost any chemical composition, such as grits, chips, lumpy materials and liquid metal, to be used as additives. This surfacing method is characterised by very high productivity — from tens to several hundreds of kilograms of metal can be deposited per hour.

The technology for ESS of large hammer and press dies using non-compacted materials (chips of tool steel and liquid filler metal) was comprehensively studied and brought to commercial application. This process involves non-consumable water-cooled electrodes and fixed moulds. As proved by industrial tests, ESS of dies enables a 2–3 times reduction in consumption of die steel and makes it possible to recycle chips to advantage directly at an enterprise that manufactures the dies.

Installations for ESS of dies were designed and manufactured. One of them, assembled in a die surfacing bay of the Open Joint Stock Company «Tokmaksk Press-Forging Factory», allows surfacing of up to 1000 tons of dies a year.

The Department also developed the method for electroslag surfacing using strips (ESSS). It is realised in flat position using one or two electrode strips. In the latter case the strips are arranged in parallel to each other and fed to the slag pool at a certain gap. Characteristic peculiarities of ESSS include high productivity (up to 60 kg/h) and very shallow penetration (the share of base metal in the deposited one is 5–15 %), which guarantees the assigned chemical composition of deposited metal in the first deposited layer.

The method is primarily employed for anti-corrosion treatment of flat parts and large-diameter bodies of revolution, using strips of stainless steels and nickel alloys. Here the use is made of cold-rolled or sintered electrode strips 30–80 mm wide and 0.5–1.0 mm thick, as well as fluxes providing high stability of the electroslag process, good detachability of the slag crust, excellent formation and high quality of deposited metal.

The Department developed technologies based on electroslag heating for making different master alloys, which are used as a charge of flux-cored wire for surfacing and arc metallising. Studies are in progress on recycling of industrial wastes using the resource-saving technologies based on the electroslag processes.

Electroslag heating served as a basis for the development of technologies for non-oxidation heat treatment of various critical parts, brass plating of metal cords and production of bimetal steel + bronze parts for hydraulic transmission boxes. To implement these technologies, associates of the Department together with the Design Bureau and Pilot Plant for Welding Equipment of the E.O. Paton Electric Welding Institute developed the corresponding equipment and introduced it at a number of industrial enterprises.

The Department conducted research in the field of other energy sources to be used for surfacing, such as electron beam, lasers and induction heating. Comprehensive studies were completed to commercialise production of wear-resistant bimetal plates and profiles (share strips, side members of coal cutter-loader chutes, etc.) by roll-welding of multi-layer stacks. Design of the stacks and the rolling technology provide wear-resistant bimetal plates with the assigned ratio of layers and cladding of the entire surface of a plate. The cladding layer in bimetal profiles is located in zones of the highest wear. The selected ratio of the layers, e.g. in share strips, provides the self-sharpening effect of the shares.

The Department is working on development of fundamentally new surfacing consumables for tribological applications, which are employed for repair and hardening of parts operating under conditions of metal on metal friction (shafts, axles, crane wheels, cold and hot forming dies, general engineering fixture components, etc.). New surfacing consumables are alloyed with elements that individually or combined with other elements act as grease, thus substantially decreasing wear and friction losses.

In the last years, the Department in collaboration with the Department for Mathematical Studies of the E.O. Paton Electric Welding Institute has developed computerised systems for design of arc surfacing technologies. One of such systems is intended for development of the technologies for surfacing of machine parts and mechanisms operating in different industries: mining, metallurgical and chemical engineering, transport, agriculture, etc. It is used to solve such critical problems related to design of the surfacing technologies as selection of surfacing consumables depending upon the service conditions and types of wear, surfacing methods and techniques, sub-layer material, surfacing parameters and conditions for preheating and subsequent cooling of workpieces.

The second system is intended for development of the technologies for surfacing of machine parts and mechanisms operating at metallurgical enterprises. The system contains information on more than 300 parts of metallurgical equipment, which can be treated by surfacing. It gives recommendations on appropriate types of surfacing consumables, surfacing conditions, heat treatment parameters, shielding atmospheres, equipment and other factors of the arc surfacing technology for each of the parts.

The Department maintains close contacts with industrial enterprises and companies in Ukraine and abroad, and performs surfacing works for them under contracts and agreements. Associates of the Department take an active part in conferences and workshops on welding and related processes and regularly publish results of their research in scientific-and-technical journals. The best developments of the Department were marked with diplomas and medals at national and international exhibitions.

SECONDARY HARDENING OF DEPOSITED METAL OF THE TYPE OF DISPERSION-HARDENING STEEL OF Fe–C–Ni–Cr–Si–Al–Cu ALLOYING SYSTEM

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Metallographic studies of deposited metal of dispersion-hardening steel of Fe–C–Ni–Cr–Si–Al–Cu system have been conducted. It is shown that secondary hardening of deposited metal proceeds in two stages, depending on tempering duration. In the first stage it proceeds through residual austenite decomposition and martensite formation. In the second stage with longer duration of tempering, martensite decomposes with formation of new phases, namely high-alloyed ferrite and dispersed inclusions with a higher content of silicon and nickel.

Keywords: *arc surfacing, surfacing materials, self-shielded flux-cored wires, dispersion hardening, structure*

In the general case hardening of dispersion-hardening steels and alloys is due to precipitation of particles of the dispersed phase through decomposition of an oversaturated solid solution. The hardening level, in particular increase of hardness, is determined by the nature of decomposition, size, shape and structure of the precipitating particles. In different steels and alloys it is manifested with certain features and dispersion hardening is found even in the most widely-used low-carbon structural steel St3 [1].

Particles of the dispersed phase, as a rule, are very thin plates or needles and during examination they create the effect of anomalous scattering of X-rays, because of their shape features. In the case of absence of evident indications of a new precipitating phase, Guinier–Preston zones can be considered as the second phase particles [2]. Formation of such zones or coherently-bonded particles of a phase with the structure in the isomorphous matrix, similar to that of other intermediate phases, is associated with the boundary of metastable equilibrium in the constitutional diagram. Thus, a continuous sequence of decomposition stages is assumed right up to precipitation of three-dimensional particles of stable and metastable phases.

As shown by analysis of published data [3–6 etc.], dispersion-hardening steels and alloys are used mainly in manufacture of cutting tools. Most of these materials are high-alloyed steels and alloys of W–Co–Fe, W–Mo–Cr–Fe, Cr–Mo–V–Fe, Ni–Cr–Fe and other systems. All of them demonstrate a pronounced effect of dispersion-hardening and after appropriate heat treatment they acquire high service properties. However, their wide application in industry is restrained by their high cost, as the content of expensive alloying elements in these steels (W, Mo, Co) is 20 wt.% and more. In addition, they, as a rule, feature a high initial hardness (*HRC* 40–55).

Rather well-studied are low- and medium-carbon dispersion-hardening Cr–Mn steels alloyed with nickel, molybdenum, vanadium and other elements, containing 2–20 wt.% Cr and 1–10 wt.% Mn [7–10]. High-alloyed dispersion-hardening steels are little studied. Based on some data [11, 12] they show a significant increase of hardness at tempering (up to *HRC* 25–30).

This work is a study of some features of secondary hardening of high-carbon (about 1 wt.% C) deposited metal of Fe–C–Ni–Cr–Si–Al–Cu alloying system. The goal was to develop a surfacing material, the hardness of which would be increased by not less than 2 times after tempering – from *HRC* 20–25 (after surfacing) to *HRC* 48–52 (after tempering).

Table 1 gives the composition of metal deposited using test flux-cored wires. Preliminary experiments demonstrated that secondary dispersion hardening of deposited metal of the selected alloying system is mainly provided by silicon, aluminium and copper. In order to study their influence on this process, the total content of these alloying elements in test flux-cored wires No. 1–3 was varied in the range of 5.0–9.5 % (Table 1). In wires No. 4 and 5 carbon weight fraction is reduced, this allowing determination of the lower limit of its content – up to 0.7 wt.%; upper limit being not more than 1.0 wt.%, as at higher content of carbon the deposited metal develops cracks. In wires No. 6–8 nickel is partially or completely replaced by a similar austenizer – manganese, to reduce their cost.

Experimental procedure. Metal for investigations was produced by surfacing 20 × 50 × 200 mm plates of steel St3 with test self-shielded flux-cored wires (see Table 1) of 2 mm diameter in four layers without preheating in the following mode: current of 200–250 A, arc voltage of 22–24 V. After surfacing the samples were cooled in air.

Samples for metallographic studies were cut out in an electric-erosion machine tool. Samples were fastened in mandrels, ground and polished by a conventional procedure applied in metallography. One part of the sections were studied in as-deposited condition,

Table 1. Composition and hardness of metal deposited with test wires

No. of flux-cored wire	Wire grade	Weight fraction of elements*, %					Hardness	
		C	Mn	Si + Al + Cu	Ni	Cr	HB after deposition	HRC after tempering at 550 °C with soaking for 6 h
1	PP-Np-DTS1	1.0	0.5	5.0	8.0	5.0	163	18–22
2	PP-Np-DTS2	1.0	0.5	7.0	8.0	5.0	197	27–29
3	PP-Np-DTS3	1.0	0.5	9.5	8.0	5.0	179	52–53
4	PP-Np-DTS4	0.7	0.5	9.5	8.0	5.0	143	50–52
5	PP-Np-DTS4	0.4	0.5	9.5	8.0	5.0	480	50–52
6	PP-Np-DTS5	1.0	2.5	9.5	5.5	5.0	217	53–54
7	PP-Np-DTS6	1.0	4.0	9.5	4.0	5.0	229	50–52
8	PP-Np-DTS7	1.0	8.0	9.5	0	5.0	207	53–55

*Design composition of the wire.

the other — after tempering at 550 °C at different soaking times.

Metallographic examination of deposited metal samples was performed in an optical microscope «Neophot 32» with 200 to 1000 magnification. Measurement of microhardness of the structural components was performed in LECO microhardness meter at 10 g load. Scanning electron microscope JSM-840 and Ferritgehaltmesser 1.053 ferritometer were also used for the above studies.

Results and their discussion. As follows from Table 1, in order to ensure secondary hardening the deposited metal composition should contain a total amount of not less than 9.5 % Al, Si and Cu. An attempt to replace expensive nickel by a similar austenizer — manganese, failed as in this case the deposited metal susceptibility to cold cracking increased.

Metal deposited with flux-cored wire No. 3 was selected to study the mechanism of secondary hardening. No cracking was found in metal of this type. One of the lowest hardness values (*HB* 179) after deposition and one of the highest values (*HRC* 52–53) after tempering were registered in it. Samples of deposited metal No. 3 were studied in the initial condition (directly after deposition) and after annealing at 500 °C with soaking for 1, 2, 3 and 6 h.

Metallographic examination demonstrated that the structure of this type of metal in the initial condition is austenite with isolated inclusions of carbides and small content of carbide eutectic, located mostly along the grain boundaries (Figure 1, *a*). Proceeding

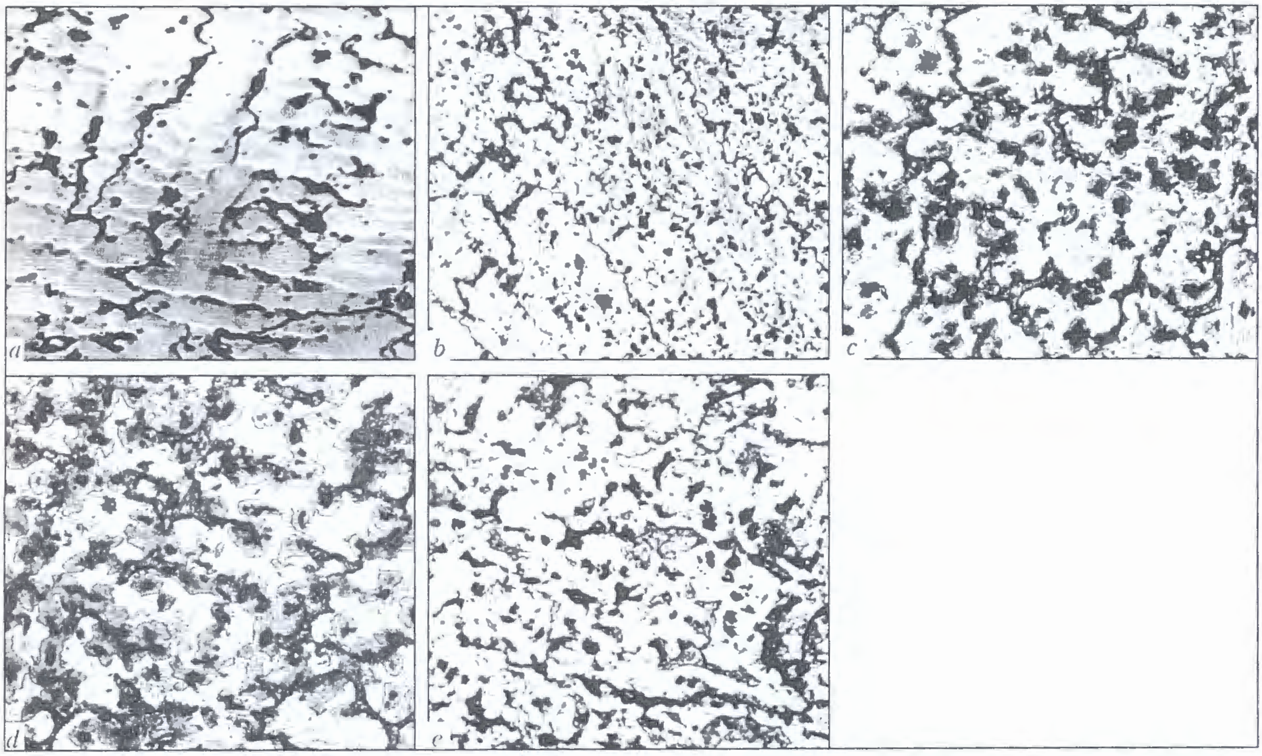

Figure 1. Microstructure of the deposited metal produced using flux-cored wire No. 3 (see Table 1): *a* — in the initial condition; *b*–*e* — after tempering at 550 °C for 1, 2, 3 and 6 h, respectively. Electrolytic etching in a water solution of chromium anhydride at *U* = 3–4 V for 20–40 s (×400)

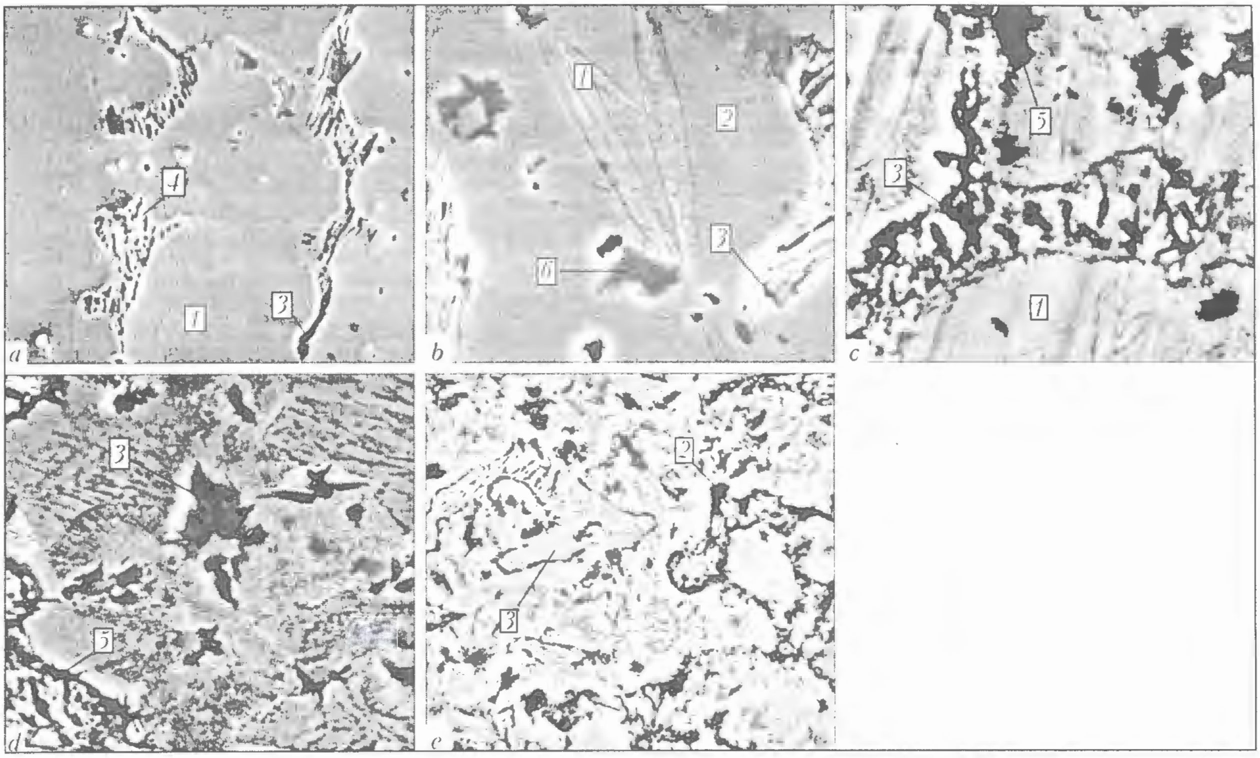

Figure 2. Microstructure of deposited metal produced using flux-cored wire No. 3 (see Table 1), and its phases, where the composition was determined by X-ray spectrum analysis (Table 2): *a*–*e* — see Figure 1; *a, e* — ×2000; *b*–*d* — ×4500; numbers indicate the studied phases

from ferritometer readings, it may be assumed that α-phase is absent in the deposited metal. Microhardness of the austenite matrix is *HV* 1880–2300 MPa (here and further on measurements were taken at 0.1 N load).

Examination in scanning electron microscope JSM-840 at 5000 magnification showed that the eutectic includes practically all the alloying elements, contained in steel (Figure 2, *a*, and Table 2, point *4*). Compared to the matrix (Figure 2, *a*, and Table 2, point *1*), the content of chromium and silicon in the carbide eutectic is higher. Located along the eutectic boundaries, are isolated dark inclusions with an increased content of silicon and nickel, compared to the matrix and eutectic (Figure 2, *a* and Table 2, point *3*).

In the case of tempering at 550 °C with soaking for 1 h, austenite decomposition occurs and a significant volume fraction of martensite forms in the metal

Table 2. Composition of deposited metal sections

Figure	*Analyzed area (point number)*	*Weight fraction of elements, %*					
		Fe	*Cr*	*Cu*	*Si*	*Ni*	*Ti*
2, *a*	Matrix (*1*)	82.69	3.51	2.94	3.02	7.56	0.29
	Inclusion (*3*)	77.64	3.05	1.88	6.80	10.18	0.44
	Eutectic (*4*)	69.97	15.22	2.37	4.78	7.69	–
2, *b*	Martensite (*1*)	83.08	3.55	2.59	3.34	7.44	–
	Austenite (*2*)	82.25	4.00	3.61	2.98	7.17	–
	Ferrite (*3*)	79.56	3.73	2.11	5.86	8.74	–
	Inclusion (*6*)	78.04	2.18	2.36	7.68	9.38	0.37
2, *c*	Austenite (*1*)	82.40	3.59	3.06	2.80	7.92	0.24
	Ferrite (*3*)	79.88	2.80	1.91	7.28	7.83	0.31
	Inclusion (*5*)	78.38	1.79	1.77	7.56	10.41	0.09
2, *d*	Ferrite (*3*)	82.25	2.31	1.63	4.25	9.18	0.38
	Inclusion (*5*)	77.83	2.53	1.75	6.61	10.77	0.51
2, *e*	Inclusion (*2*)	77.75	2.75	0.72	11.28	8.36	1.07
	Inclusion (*3*)	77.87	3.50	1.28	9.05	9.15	0.60

structure, this, apparently, determining the first stage of ageing. This results in the hardness increasing up to *HRC* 40–44. Martensite has an acicular structure with predominant needle orientation at an angle of 60° to each other. In some needles a central shear plane — midrib, is observed. The deposited metal structure also has inclusions of carbides and carbide eutectic (see Figure 1, *b*). Martensite microhardness is *HV* 3980–4900 MPa.

At a large magnification, a new phase is detected in the deposited metal structure, which was initially identified as ferrite. A higher silicon content was registered in this phase. Similar to the initial condition, inclusions with a higher content of silicon and nickel are also present in the location of the eutectic component (Figure 2, *b* and Table 2, point *6*). Content of the main alloying elements in martensite needles and residual austenite is approximately the same (Figure 2, *b* and Table 2, points *1, 2*).

Proceeding from ferritometer readings, total content of α-phase in the deposited metal structures is higher than 55 vol.%.

After soaking for 2 h at tempering temperature martensite decomposition and precipitation of carbide inclusions are observed. Amount of carbide eutectic is increased (see Figure 1, *c*). After soaking for 2 h at 550 °C macrohardness rises up to *HRC* 54–56.

Composition of the steel matrix remains unchanged (Figure 2, *c* and Table 2, point *1*). In the new α-phase silicon weight fraction rises up to 7.28 % (Figure 2, *c* and Table 2, point *3*). The process of formation of a phase with a higher content of nickel and silicon proceeds in the eutectic regions (Figure 2, *c* and Table 2, point *5*). These changes are accompanied by further increase of metal microhardness: *HV* 4900–6500 MPa.

In case of tempering with soaking for 3 h, further decomposition of martensite and increase of alloyed ferrite content take place. Content of carbides and carbide eutectic rises along the grain boundaries and in the grains proper (see Figure 1, *d*). Increase of the time of soaking at tempering temperature up to 3 h does not affect either the hardness, which remains on the level of *HRC* 54–56, nor the microhardness, the maximum value of which still is *HV* 6500 MPa.

Composition of the main structural components practically does not change. The highest content of nickel is found in ferrite and the dark phase located in the eutectic (Figure 2, *d* and Table 2, points *3* and *5*). An increased content of silicon — 6.61 wt.% is recorded in the dark component of the eutectic phase (Figure 2, *d* and Table 2, point *5*).

After soaking at the temperature of 550 °C for 6 h, the deposited metal hardness is somewhat lowered (*HRC* 52–53), microhardness also decreases and does not exceed *HV* 6060 MPa. Martensite decomposes completely, and partial dissolution of carbides and carbide eutectic is observed (see Figure 1, *e*). The process of coagulation of the inclusions in alloyed ferrite, the content of which exceeds 55 wt.% by ferritometer readings, becomes much more pronounced. In the eutectic dark component silicon content (9.05–11.28 wt.%) is increased even more at a rather high content of nickel and lower content of copper (Figure 2, *e* and Table 2, points *2* and *3*).

Thus, the mechanism of secondary hardening of the studied deposited metal at tempering can be represented as a multi-stage process.

In the first stage at soaking at tempering temperature for 1 h formation of martensite structure (pre-dispersion hardening) is observed, as well as appearance of hardening inclusions, namely a new α-phase — high-alloyed ferrite with a higher content of silicon or silicon and nickel, as well as dark component in the carbide eutectic with a higher content of silicon and nickel.

In the second stage, at increase of soaking time to 2–3 h, martensite decomposes, carbides and carbide eutectic appear in the deposited metal structure, volume fraction of the new hardening α-phase becomes greater. Silicon and nickel content in it and in the dark inclusions in the carbide eutectic becomes greater. In the case of soaking at the temperature of 550 °C for 6 h, growth of α-phase proceeds simultaneously with an opposite process in terms of hardness change, which is accompanied by coagulation of inclusions of α-phase and increase of silicon content in it above 10 wt.% and that of nickel above 8 wt.%. At such a content of these elements α-phase becomes «softer» and secondary hardness of the deposited metal is somewhat reduced, respectively.

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BEHAVIOR OF PARTICLES OF A NON-COMPACT FILLER AT AIR–SLAG INTERFACE IN ELECTROSLAG HARDFACING

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Data are given on behavior of particles of non-compact materials (chips and shot) on the surface of the pool in electroslag surfacing. Conditions have been determined for fixation of these particles at the air–slag interface. It is shown that shot of chromium cast iron with a diameter of up to 2.3 mm and chips of tool steel 5KhNM up to 2.7 mm thick (fixation by the largest side) or up to 0.6 mm thick (fixation by the smallest side) may float on the surface of the slag pool.

Keywords: *electroslag hardfacing, non-compact materials, slag pool, metal pool, interface*

Non-compact materials (NCM) in the form of shot [1, 2], chips [2–4], wire cuts [1, 2] or other particles [3, 5] are used for a long time in electroslag processes, in particular electroslag hardfacing (ESH).

The ESH NCM is peculiar by the fact that NCM particles can stay on the surface of a slag pool for some time, not immersing into it [1, 3, 4, 6]. From the one hand, it may lead to their oxidation and disturbing of the electroslag process [3, 6], while, from the other hand, the heating of particles at the slag pool surface promotes the rapid melting of skull forming from them [7] that reduces the risk of deposited metal contamination with slag inclusions and formation of other defects.

In literature this problem was not almost reflected. Existing publications are not numerous and refer to the case of floating of molten metal drops [8] and particles of copper matte [9] on the slag surface, and also to floating of metal bodies in iron-carbon melts [10]. It should be noted that assumption was taken in works [8, 10] about the absolute non-wetting of floating particles, that was not observed in real processes [11, 12].

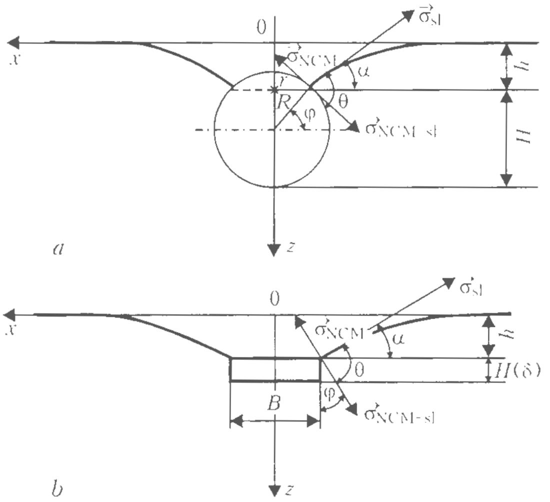

Figure 1. Scheme of arrangement NCM particles on the slag pool surface and direction of vectors of surface tension at the phase interface: *a* – sphere; *b* – plate; $\vec{\sigma}_{NCM}$ – surface tension at NCM–air interface; $\vec{\sigma}_{NCM\text{-}sl}$ – surface tension at NCM–slag interface (the rest designations see in the text)

The aim of the present work was to conduct the analytical evaluation of behavior of NCM particles on the slag pool surface at the conditions close to those of real electroslag processes. Analysis of behavior of NCM particles on the slag pool surface was made relative to particles in the form of a sphere (analogue to shot) and rectangular plate (chips), used most often in ESH NCM [2, 5]. The method offered can be used for evaluation of behavior of particle of another shape.

As was above-mentioned, the NCM particles, entered the air–slag melt interface, can float for some time at its surface. At this moment a complex thermal and hydrodynamic interaction of particles with a slag melt is occurred, consisting in a non-stationary movement of particles in a slag, formation of skull at their surface and its subsequent melting, in heating of a particle by the molten slag heat. In this case the density of a particle–skull system and wetting angle are continuously changed.

The present work considers particles which are not moved along the surface of the slag pool, but located at its surface until complete melting of the skull. At the temperature of skull melting the angle θ of wetting the metal solid surface with slag will be less than $\pi/2$ [12, 13].

As the density of NCM particles ρ_{NCM} is higher than slag density ρ_{sl}, then they can be maintained at the slag melt surface only by the forces of surface tension [11, 13]. Here, the vertical constituent of forces of surface tension (Figure 1) is equal to [11]

$$F_{\sigma z} = l\sigma_{sl}\sin\alpha, \qquad (1)$$

where l is the length of wetting perimeter; σ_{sl} is the surface tension of slag; $\alpha = \theta + \varphi - \pi/2$ is the angle of inclination of slag surface to horizon at wetting perimeter; θ is the hysteresis (or equilibrium in particular case) angle of wetting; φ is the angle of particle shape.

As is stated in works [1, 10], the nature of wetting has a great influence on maintaining of solid bodies of different shape on the surface of melts. It is outlined in the mentioned works that maintaining of solid plates on the melt surfaces is possible only in case of a negative wetting ($\theta > \pi/2$). However, this problem is of a discussion nature. We have found in conducting experiments on remelting of chips of die steels (DI-22, 5KhNM) [4] and waste of saw blades (steel R3M3F2) [5] in slag of CaF_2–Al_2O_3–CaO–SiO_2, that particles in the form of plates continue to float in the slag melt even in heating of their surface up to solidus temperature of the material remelted. Here, the positive wetting is typical ($\theta < \pi/2$).

On the basis of data of works [13, 14] the process of fixation of particles, having plane facets, at the surface of the slag pool at $\theta < \pi/2$ can be presented as follows (Figure 2).

At the beginning, when particle NCM enters the surface of the slag pool, the vector of surface tension σ_{sl}, directed downward under angle θ to the lateral facet of particle, promotes its immersion. As a result, the particle begins to immerse into the melt and wetting perimeter is moved from its lateral facets to the zone of upper edge (regions of the particle surface, locating from edge at the distance equal to the radius of action of molecular forces). Here, vector σ_{sl} changes the direction and at $\theta + \varphi > \pi/2$ will be directed upward under some angle α to horizon, promoting the fixation of particles at the slag pool surface.

Transition of wetting perimeter through the zone of edge can be considered as the movement along the curvilinear surface of a very small radius of curvature, commensurable with a radius of action of molecular forces. Here, angle α and the force $F_{\sigma z}$, determined by it, will increase. At the moment of particle diving the wetting perimeter will pass in the zone of edges along the upper facet, due to which angle φ will become equal to $\pi/2$, and the angle α will reach its maximum value, i.e. hysteresis angle of wetting θ.

Hence, it follows that fixation of particles with plane facets at the slag melt surface can occur at sufficiently small angles of contact. For particles, entered the air–slag interface under angle or by a small facet, the probability of fixation at the interface is lower than for the particles falling on the interface by one of large facets.

For the bodies with a curvilinear surface the direction of surface forces is also changed in immersion and promotes their forcing out at $\theta < \pi/2$ [15]. However, the conditions of fixation of these particles at the interface are less favorable as in this case the maximum angle φ is always less than $\pi/2$. The position of particle at the interface (see Figure 1) is defined by angle φ characterizing the position of wetting parameter on the particle surface and the height h equal to the shifting of slag surface in vertical in the point 0.

The analysis is aimed at the determination of dependence of angle φ (or α) on radius R (or thickness

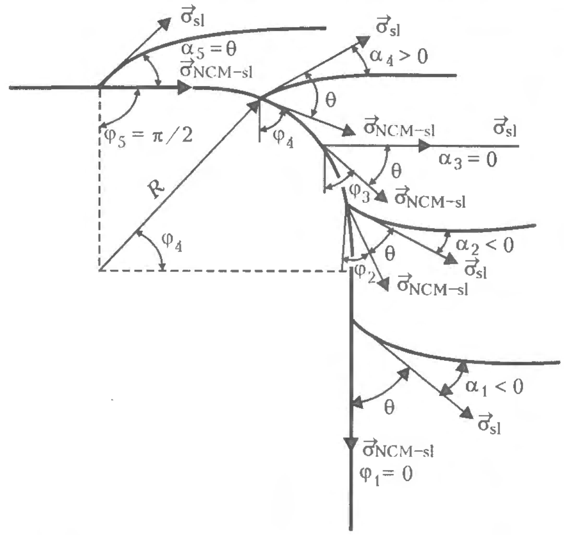

Figure 2. Displacement of wetting perimeter through a zone of edges of NCM particles, having plane facets (α_1–α_5, φ_1–φ_5 – changes of angles α and φ in immersion of NCM particle into slag)

δ) of particle and contact angle θ, and also maximum sizes of floating particles.

Sphere. The following forces are acting on the sphere of radius R, floating in slag:
gravity:

$$\mathbf{P} = V\rho_{NCM}\mathbf{g} = \frac{4}{3}\pi R^3\rho_{NCM}\mathbf{g};$$

Archimedes:

$$\mathbf{F}_A = V_s\rho_{sl}\mathbf{g} = \pi H^2\left(R - \frac{H}{3}\right)\rho_{sl}\mathbf{g} =$$

$$= \frac{1}{3}\pi R^3(1 + \sin\varphi)^2 (2 - \sin\varphi)\rho_{sl}\mathbf{g};$$

hydrostatic pressure (part of Archimedes' force defined by the difference of levels of wetting perimeter and slag surface):

$$\mathbf{F}_p = \pi r^2 h\rho_{sl}\mathbf{g} = \pi R^2 h\rho_{sl}\mathbf{g}\cos^2\varphi;$$

surface tension:

$$\mathbf{F}_\sigma = 2\pi r\sigma_{sl} = 2\pi R\sigma_{sl}\cos\varphi,$$

where V is the sphere volume; V_s and $H = R(1 + \sin\varphi)$ are, respectively, volume and height of immersed segment; $r = R\cos\varphi$ is the radius of its base; g is the acceleration of a free fall.

At $R << a$, that corresponds to the radius of shot used for ESH NCM, it is possible to determine value h, using expression given in work [15]:

$$h = -R\cos\varphi\,\mathrm{tg}\,\alpha\left(\ln\frac{R\cos\varphi}{2a} + \gamma\right),$$

where $a^2 = \sigma_{sl}/\rho_{sl}g$ is the capillary constant of slag; $\gamma = 0.5772$ is the Euler's constant.

Condition of equilibrium of particles at the air–slag melt interface is the equality to zero of value of sum of projections of acting forces on vertical axis. In this case, if the constituent of force vector is directed upward, then the projection is considered positive, and if it is directed downward, then the projection is considered negative. Thus, after proper transformations it can be written

$$R^2[(1+\sin\varphi)^2(2-\sin\varphi) + 3\left(\ln\frac{R\cos\varphi}{2a}+\gamma\right)\cos^3\varphi \times \operatorname{ctg}(\theta+\varphi) - 4\frac{\rho_{NCM}}{\rho_{sl}}] - 6a^2\cos\varphi\cos(\theta+\varphi) = 0. \quad (1)$$

Solution of transcendental equation (1) relative to φ cannot be obtained in a clear form, however, the determination of R at different φ is not difficult.

Value of angle φ_{max} corresponds to maximum size of particles which are floating on the slag surface. For its finding it is necessary to equalize the derivative $d(\Sigma F_z)/d\varphi$ to zero. It is difficult to determine the value φ_{max} analytically from this condition. It should be noted that extreme values ΣF_z and $F_{\sigma z}$ are almost at the same values φ, as the surface tension force undergoes the largest change in particle diving both in value and in direction. Therefore, the necessary value will be defined from the condition $dF_{\sigma z}/d\varphi = 0$ at $d^2F_{\sigma z}/d\varphi^2 < 0$. Hence

$$\varphi_{max} = \frac{\pi-\theta}{2}.$$

Substituting value φ_{max} into equation (1), we shall find expression for determination of a maximum radius of particles of a spherical shape, maintained at the slag surface:

$$R^2_{max}\left[\left(1+\cos\frac{\theta}{2}\right)^2\left(2-\cos\frac{\theta}{2}\right) - 3\left(\ln\frac{R_{max}\sin\theta/2}{2a}+\gamma\right)\times \sin^3\frac{\theta}{2}\operatorname{tg}\frac{\theta}{2} - \frac{4\rho_{NCM}}{\rho_{sl}}\right] + 6a^2\sin^2\frac{\theta}{2} = 0. \quad (2)$$

Thus, the value R_{max} depends on the slag surface tension, wetting angle and density of contacting phases.

Plate. The plate of thickness δ, floating in slag, is subjected to the action of the force of gravity, forcing out, hydrostatic pressure and surface tension, having the following expressions:

$$\mathbf{P} = V_p\rho_{NCM}\mathbf{g} = \delta BL\rho_{NCM}\mathbf{g};\quad \mathbf{F_A} = V_p\rho_{sl}\mathbf{g} = \delta BL\rho_{sl}\mathbf{g};$$
$$\mathbf{F}_p = Sh\rho_{sl}\mathbf{g} = BLh\rho_{sl}\mathbf{g};\quad \mathbf{F}_\sigma = 2(B+L)\sigma_{sl},$$

where V_p is the plate volume; S is the base area; B is the width; L is the length.

Conditions of plate equilibrium at the slag surface have the form

$$\delta BL\rho_{sl}g + BLh\rho_{sl}g + 2(B+L)\times \sigma_{sl}\sin\alpha - \delta BL\rho_{NCM}g = 0. \quad (3)$$

In a general case for particles, similar in shape, $ah = f(L, B, \alpha)$ at a preset value.

If the metal remelted has $B \neq L$ and $B/2 >> a$, that is typical of the main mass of chips used in ESH NCM [4], then h can be determined using formula (2) given in work [10]:

$$h = 2a\sin\frac{\alpha}{2}.$$

Then, the expression (3), after proper transformations, will take the form

$$\delta = \frac{2a\sin\frac{\alpha}{2}\left(ma\cos\frac{\alpha}{2}+1\right)}{\rho_{NCM}/\rho_{sl} - 1}, \quad (4)$$

where $m = 2(B+L)/BL$ is the ratio of plate perimeter to base area. At $\alpha = \alpha_{max} = \theta$ it is possible to determine the maximum thickness of plate maintained by the slag using formula (4).

If the material remelted has $L >> B$, then it is convenient to consider plate of infinite length. In this case the expression (4) will have $m = 2/B$.

To check the validity of obtained relationships, let us consider several definite examples of calculation for electroslag technologies used in industry where NCM are used.

Electroslag hardfacing of hot rolling mill rolls using a grained filler material [6]. Hardfacing material is a cast shot of high-chromium cast iron of density $\rho = 7600$ kg/m^3 [16]. Slag AN-75 [17] is used for hardfacing, which is close by composition to slag ANF-14. According to work [18], at 1460 °C (temperature of slag surface at the given method of hardfacing [6]) slag ANF-14 has properties (ρ_{sl} = 2650 kg/m^3 and σ_{sl} = 0.31 N/m), which can refer at some approximation to slag AN-75. The contact angle of wetting the surface of high-chromium cast iron with slag AN-75 at 1300 °C (solidus temperature of cast iron) reaches 42.5° [17].

Results of calculation of values R for different angles φ by formula (1), and also R_{max} by expression (2) are given in Figure 3. According to data of work [6], the shot of diameter of not more than 1.5 mm (or $R \leq 0.75\cdot10^{-3}$ m) is floating stable at the slag pool surface, that has a good correlation with calculated data.

Electroslag hardfacing of worn-out dies using chips [5]. Hardfacing material is chips of die steel

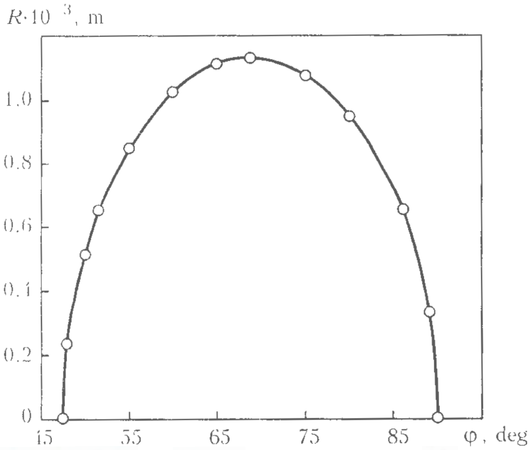

Figure 3. Dependence of radius of cast iron shot floating on the slag surface, on shape angle of particle φ

5KhNM of density ρ_{NCM} = 7849 kg/m^3 [19]. There are no data in literature about ρ_{sl} and σ_{sl} for slag AN-15M used in hardfacing of dies, therefore, we shall use data of work [20] for slag, close by composition to AN-15M, wt %: 25CaF_2; 33.65Al_2O_3; 30.85CaO; 10.5SiO_2. At 1600 °C ρ_{sl} = 2660 kg/m^3, σ_{sl} = 0.39 N/m are for this slag. There are also no data about contact angle of wetting of surface of steel 5KhNM with slag AN-15M. According to data of work [17], at 1300 °C temperature the angle of wetting of surface of steel 45 with slag ANF-29, close by composition to slag AN-15M, reaches 38.5°. At 1460 °C (solidus temperature of steel 5KhNM [19]) angle θ, equal to 40°, is taken for calculation.

In hardfacing the chips and shavings from milling and planning machines are used. The former has a form of bent plates of the following sizes: δ = = $(0.3-1.0)\cdot10^{-3}$ m, $B = (8.5-11.0)\cdot10^{-3}$ m, L = (16–20)$\cdot10^{-3}$ m. When these chips enter the slag by the largest facet, the calculations from formula of [17] give the following values for $\delta_{max} = (2.743-3.130)\cdot10^{-3}$ m.

In case of chips enter by the smallest facet (i.e. edges of length L are located vertically), σ_{max} = = $(0.555-0.730)\cdot10^{-3}$ m.

Shavings of planning machines have a form of bent plates of sizes $\delta = (1.0-2.5)\cdot10^{-3}$ m, B = (11–13)$\cdot10^{-3}$ m, $L = (20-50)\cdot10^{-3}$ m or spirals of length $L = (120-150)\cdot10^{-3}$ m at similar δ and B. In case of enter of shavings in the form of bent plates by the largest facet, $\delta_{max} = (2.310-2.743)\cdot10^{-3}$ m according to calculations using expression from work [17]. In case of shavings enter by the smallest facet, δ_{max} = = $(0.206-0.555)\cdot10^{-3}$ m.

Probability of fixation of shaving in the form of a flat spiral is higher if it contacts the slag surface in all length L (i.e. edges of length B are located vertically). For the given case (plate of a finite length, as $L >> \delta$), $\delta_{max} = (0.845-1.021)\cdot10^{-3}$ m.

Thus, the calculations showed that during hardfacing the chips from milling machines are located in principle at the slag surface until its partial or complete melting, and the most part of shavings (in particular spiral-like) from planning machines can immerse and be melted under the layer of slag.

Figure 4 gives, as an example, the calculated values of thickness of floating chips (in the form of bent plate) of sizes $B = 11\cdot10^{-3}$ m, $L = 20\cdot10^{-3}$ m depending on angle φ. For comparison, the data of calculation of characteristics of equilibrium position of shot of steel 5KhNM on the surface of slag AN-15M.

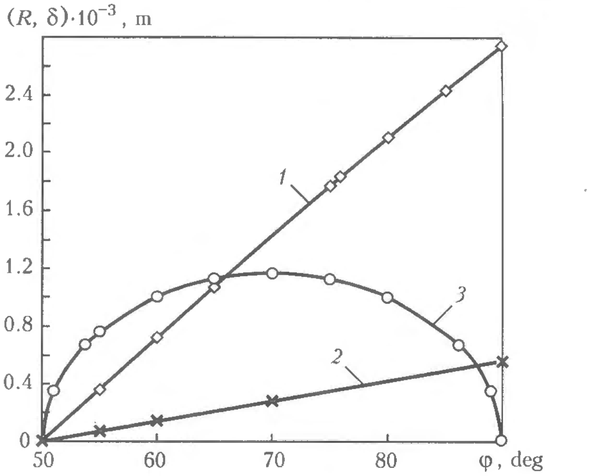

Figure 4. Dependence of sizes of NCM particles of steel 5KhNM, floating on slag surface, on angle of shape φ: *1, 2* — shavings fixed, respectively, by the largest and smallest facet; *3* — shot

CONCLUSIONS

1. Relationships describing the conditions of staying of NCM of different shape on the slag surface are obtained as a result of the theoretical analysis.

2. It was established on the basis of calculations obtained from relationships as regards to solid metal–slag melt system that chromium cast iron shot of up to 2.3 mm diameter and tool steel 5KhNM chips of up to 2.7 mm thickness (fixation by the largest facet) or up to 0.6 mm (fixation by the smallest facet) can be maintained on the slag pool surface. In this case a good correlation of experimental data with calculated data was observed.

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STUDY OF PROPERTIES OF DEPOSITED METAL OF THE MARAGING STEEL TYPE

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Structure and properties of maraging steels deposited with flux-cored wires of different alloying systems have been studied. Optimal composition of deposited metal has been determined. In the as-deposited condition this metal has hardness *HRC* 30, which makes it easy to cut, whereas after tempering at 480 °C for 3 h its hardness grows to *HRC* 50. The deposited metal after tempering has good high-temperature strength, which allows it to be recommended for hardening of working surfaces of complex-configuration die tools.

Keywords: *arc cladding, deposited metal, maraging steels, flux-cored wire, heat resistance*

Maraging steels in their operational, mechanical and service properties are very promising wide-application materials. Owing to a specific hardening mechanism, the technology of manufacture of most diverse parts from these steels is characterised by relative simplicity and reliability [1, 2].

Maraging steels used as deposited metal have a number of advantages over metal of the martensitic grade, such as the possibility of cladding without preheating and concurrent heating, comparatively low initial hardness allowing machining of treated parts by cutting, and high performance provided by tempering after machining.

The information on maraging steels available in technical literature concerns mainly low-carbon high-nickel steel alloyed with molybdenum and cobalt. High-strength steels based on the Fe–Ni–Co–Mo alloying system are classed as a rule by their tensile strength [1]. These steels are characterised by sufficiently high hardness and strength. For example, steels containing about 15 wt.% Co and 10 wt.% Mo, subjected to ageing, have hardness *HRC* 60. The application field of such steels includes ship building, rocket, aircraft and cryogenic engineering.

Several grades of flux-cored wires are available for cladding with such steels (N8M11K10ST, N12M8K8S2T, etc.) by the submerged-arc method [3, 4]. However, very high costs and short supply of alloying elements limit the possibility of wide application of such materials for hardening different parts and tools. Also, one should bear in mind that the high degree of alloying of high-strength maraging steels (the total content of alloying elements in them should be no more less 30–35 %) makes it almost impossible to manufacture flux-cored wire, especially self-shielded one, to provide this class of deposited metal.

In this connection, worthy of attention are sparsely-alloyed structural or tool maraging steels [2, 4]. The latter can be used, in particular, for cladding of die tools operating in contact with hot metal.

High strength of this class of maraging steels is a result of collective realisation of two hardening processes, i.e. formation of substitution solid solution and shear (martensite) mechanism of $\gamma \rightarrow \alpha$ transformation.

In the case of using this type of steels as cladding materials for reconditioning of die tools, they should have high heat resistance and thermal stability. To possess these properties, the steels should contain nickel, molybdenum, titanium, aluminium and other alloying elements, which provide decrease in the $\gamma \rightarrow \alpha$ transformation temperature, as well as formation of substitution martensite and hardening phases. Also, addition of 0.5–2.5 wt.% Mo to the steel leads to substantial increase in its ductility and toughness. Small addition of silicon to nickel steel decreases solubility of molybdenum and titanium, which enhances the effect of age-hardening [2].

It should be noted that the use of carbon, the content of which in such steels may amount to 0.06-0.10 wt.%, is indicated to decrease the temperature of beginning of martensitic transformations and strengthen martensite after quenching. The possibility of increasing the carbon content of steel to 0.10 wt.% is important in terms of development of cladding wires, which are manufactured from a cold-rolled strip of steel 08kp (rimmed) and C-containing ferroalloys.

Eleven experimental flux-cored wires with a diameter of 2 mm, providing metal of the type of maraging steels with different alloying systems, were made to select optimal composition of deposited metal. Cladding using all flux-cored wires was performed with no preheating under the following conditions: I_c = 240–260 A and U_a = 22–24 V. Chemical composition of deposited metal (in the fourth layer) and its hardness in the as-deposited condition and after tempering are given in Table 1.

As to their composition, experimental flux-cored wires can be subdivided into several groups (Table 1). Flux-cored wires of the first group (1–3) contain nickel, chromium and molybdenum in their charge. Chromium is absent in the charge of flux-cored wires of the second group (4–6) and nickel here is fully replaced by manganese. Flux-cored wires of the third

Table 1. Chemical composition of deposited metal

Flux-cored wire and No. grade	*Content of elements, wt.%*						*HRC*	
	C	*Ni*	*Mn*	*Mo+Ti+W*	*Si+Al+Cu*	*Other*	*As-deposited*	*After tempering*
PP-Np-Exp1	0.10	11.0	–	3.4	2.8	4.8Cr; 0.2V	28–31	43–45
PP-Np-Exp2	0.05	8.9	2.3	2.3	1.1	5Cr	29–31	40–41
PP-Np-Exp3	0.06	5.2	0.7	3.2	1.7	3.9Cr; 0.8Nb	38–39	46–48
PP-Np-Exp4	0.10	–	5.1	4.6	4.1	–	32–34	52–53
PP-Np-Exp5	0.07	–	5.1	3.7	2.0	–	20–22	34–35
PP-Np-Exp6	0.11	–	5.5	4.2	3.4	–	25–26	34–35
PP-Np-Exp7	0.07	2.3	5.2	4.0	4.6	–	39–40	44–45
PP-Np-Exp8	0.07	7.5	5.2	2.3	2.4	–	32–34	52–53
PP-Np-Exp9	0.08	8.0	5.5	2.4	1.5	–	29–30	49–50
PP-Np-Exp10	0.07	8.5	5.8	2.5	0.7	–	25–26	46–47
PP-Np-Exp11	0.08	3.9	4.1	4.2	4.5	–	34–35	48–49

Note. Ageing temperature 480–500 °C, holding for 3 h.

group (7–11) contain no chromium either, and nickel in them is partially replaced by manganese.

It was found during the deposition process that all flux-cored wires under consideration were characterised by the required welding-operational properties, formation of sound deposited metal and good detachability of slag crust (gas- and slag-forming system of wires — $CaCO_3$–TiO–CaF_2). However, multilayer cladding on a restraint massive sample resulted in small cracks formed in the Ni-free deposited metal (flux-cored wire 6) and deposited metal with 2 wt.% Ni (flux-cored wire 7). No such defects were detected in metal deposited with flux-cored wires of other groups.

Selection of optimal composition of deposited metal was based on the following requirements: hardness of the as-deposited metal should not exceed *HRC* 30, and after ageing it should be at a level of *HRC* 50. These requirements are best met by the Ni–Mn deposited metal (flux-cored wires 8 and 9) additionally alloyed with silicon, which somewhat enhances the effect of age-hardening [2]. This metal is more economic as to the alloying degree than the known maraging steels containing 12–18 wt.% Ni and up to 10 wt.% of Mo, Co, W and other alloying elements [1, 2].

Table 2. Dependence of hardness of metal deposited with flux-cored wire PP-Np-Exp9 upon the temperature and time of tempering

Tempering		*HRC*	
Temperature, °C	*Time, h*	*As-deposited*	*After tempering*
400	3	29–30	48–49
400	5	29–30	48–49
500	3	29–30	49–50
500	5	29–30	46–47
600	3	29–30	30–31
600	5	29–30	28–29

Out of the two compositions of deposited metal (flux-cored wires 8 and 9), the preference should be given to the second one, which in the as-deposited condition has hardness *HRC* 30, thus causing no problems in its machining by cutting. After tempering, hardness of this deposited metal grows to *HRC* 50. Therefore, this type of metal was chosen as the basic one for the development of self-shielded flux-cored wire and technology for deposition of a sparsely-alloyed magaring steel layer.

Dependence of deposited metal upon the temperature and time of tempering was studied to select optimal heat treatment parameters for metal deposited with flux-cored wire of the PP-Np-Exp9 grade. The results obtained are given in Table 2.

On this basis, heating to 480–500 °C and holding for 3 h should be taken as optimal parameters of tempering of metal deposited with wire PP-Np-Exp9.

Studies were conducted to investigate heat resistance (resistance to tempering) of metal deposited with flux-cored wire PP-Np-Exp9, which is the wire of choice for most applications involving repair and hardening of hot deformation tools. Normally, heat resistance is characterised by the tempering temperature at which the metal retains hardness at a level of

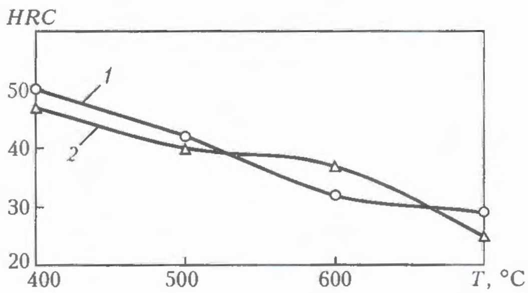

Figure 1. Heat resistance of metal deposited with flux-cored wire PP-Np-Exp9 (*1*) and die steel 5KhNM (*2*)

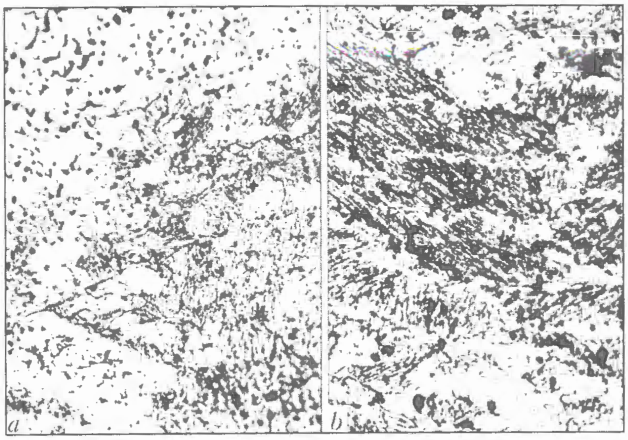

Figure 2. Microstructure of metal deposited with wire PP-Np-Exp9 in the as-deposited state (*a* — ×320) and after tempering (*b* — ×500) at 480 °C for 3 h

HRC 40. The test results are shown in Figure 1. Deposited metal of the suggested composition is not inferior in this property to known die steel 5KhNM [5].

Experiments were conducted to evaluate the effect of the cladding technology on initial hardness of metal deposited with wire PP-Np-Exp9, as well as the effect of ageing. It was shown that preheating (200–400 °C) of a part treated has no effect on initial hardness of the deposited layer and its hardness after tempering.

The cladding method (open-arc or submerged-arc) and the rate of cooling after cladding, either delayed cooling (thermostat, furnace) or cooling in open air, also have no effect on hardness of metal deposited with wire PP-Np-Exp9. An insignificant increase (2–3 Rockwell numbers) in the initial hardness of metal produced by cladding with cooling of each deposited layer to 20 °C prior to deposition of the next one was noted. In this case, hardness after tempering remained at the previous layer, i.e. *HRC* 50.

As shown by metallography, in the as-deposited condition the structure of metal deposited with wire PP-Np-Exp9 consists of austenite, ferrite and a small volume fraction of martensite (Figure 2, *a*). After tempering at 480 °C for 3 h the structure of this type of deposited metal is a low-projection laminated martensite with finely dispersed carbide inclusions (Figure 2, *b*).

Table 3. Chemical composition of structural components of metal deposited with wire PP-Np-Exp9

Analysis place	*Content of elements, wt.%*						
	Fe	*Ni*	*Mn*	*Si*	*Al*	*Mo*	*Ti*
Matrix	84.92	7.21	4.35	1.29	0.79	0.70	0.78
Inclusion	63.15	5.09	2.87	0.80	0.65	0.68	26.75

The composition of matrix and dispersed inclusions in metal deposited with wire PP-Np-Exp9 (Table 3), as well as the distribution of main alloying elements in it, were examined using X-ray microanalyser Camebax SX50.

Inclusions are complex carbides, ranging from 2 to 5 μm in size. As far as the distribution of alloying elements in deposited metal is concerned, the most uniform distribution in matrix was noted for aluminium, whereas titanium is contained mainly in carbides.

Investigation of the effect of ageing in the process of tempering of deposited metal of the type of maraging steels with different alloying systems allowed determination of the optimal composition of self-shielded flux-cored wire PP-Np-Exp9. The hardness of metal deposited with this wire was found to be independent of the cladding method and temperature cycle. Also, this type of deposited metal has sufficiently high heat resistance, which allows it to be recommended for hardening of working surface of complex-configuration dies.

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EFFECT OF INDUCTION SURFACING CONDITIONS ON STRUCTURE AND PROPERTIES OF DEPOSITED METAL

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The paper gives results of investigation of structure and properties of metal deposited by the induction surfacing method that involves screening of thermal and electromagnetic fields. It is shown that owing to screening the non-uniformity of thickness of the deposited metal layer is decreased by 12 %, saving of power amounts to 10–15 %, overheating of the edge of disc and deposited metal is eliminated, and the time of surfacing is reduced as compared with surfacing without screening. In this case the microstructure, hardness and wear resistance of the deposited metal remain at the same level as in induction surfacing without screening.

Keywords: *induction surfacing, inductor, specific power, thermal and electromagnetic screening, deposited metal, carbides, microhardness, structure, wear resistance*

To save the electric power in induction surfacing of knives of beet-topping machines, the authors of the present work have suggested to change the heat source power by the exponential law (i.e. at the continuous increment of specific power) without the generator switching [1]. In this case the 15–25 % of power saving is attained. This idea was realized in the development of technology of the induction surfacing of knives of beet-topping machines with screening of thermal and electromagnetic fields [2, 3].

The aim of the present work was to investigate the effect of induction surfacing conditions on structure and properties of metal, deposited by induction method using screening of thermal and electromagnetic fields, and also to compare them with similar properties of metal deposited by conventional induction method without screening.

Surfacing by two technologies was performed using the wear-resistant powdered alloy PG-S1 (U30Kh28N4S4). Thickness of deposited layer was 0.8–1.5 mm, the parent metal was steel St3 of 3 mm thickness. Two-turn circular inductor, equipped with auxiliary devices for screening of thermal and electromagnetic fields, was used.

As the investigations showed, the microstructure of metal deposited by the technology, developed by the authors, contains primary chromium carbides in the form of rather large plates of a rectangular or rhombic shape, which are distributed uniformly in the matrix (Figure 1). Clusters of carbide eutectic are adjacent to the interface from the side of deposited metal. At the fusion line, the deposited metal has a clearly expressed dendritic structure (Figure 1, *a*). The inclusions of primary carbides of a fan-like shape are contained in the upper layer of the deposited metal (Figure 1, *b*). White band at the interface between the parent and deposited metal has a variable width (15–20 μm). Along the edges of the deposited bead the structure represents a mixture of carbides of different dispersity. In this case the presence of primary plate carbides is not observed.

Using the LECO microhardness meter M-400 (Table 1) the microhardness (MPa) of structural constituents of metal deposited without and with screening of thermal and electromagnetic fields was measured.

As is seen from the data obtained, the values of microhardness of structural constituents of metal deposited by two technologies are not almost differed.

X-ray microanalysis of the deposited metal was made in the Cameca microanalyzer Camebax SX-50 (Figure 2). In all the cases the analysis was performed approximately in the center of the deposited layer normal to the fusion line at the depth up to 480 μm

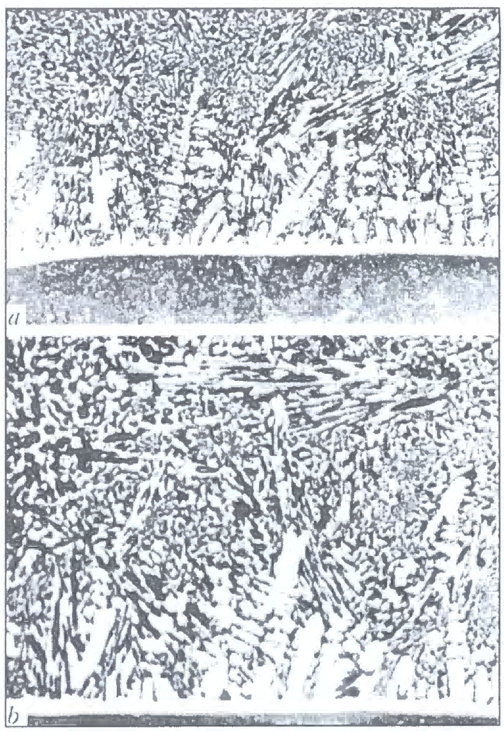


Figure 1. Microstructure of deposited metal U30Kh28N4S4 obtained directly at the fusion line (*a*) and in the center of deposited layer (*b*); electrolytic etching in 20 % water solution of chromium anhydride, 20 V voltage, 5 s holding (×200)

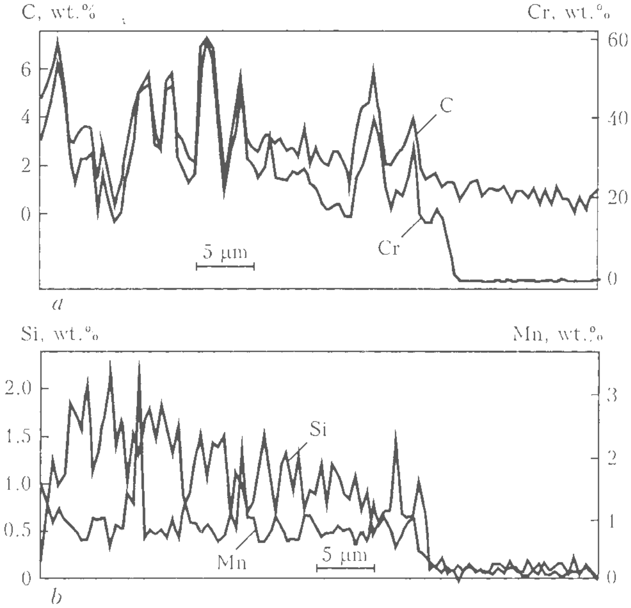


Figure 2. Distribution of carbon and chromium (*a*), silicon and manganese (*b*) in height of deposited metal layer

from the fusion line. It was established that carbon in metal of samples examined was bound into carbides of Me_7C_3 type (Figure 2, *a*), the noticeable diffusion redistribution of carbon near the fusion line was not observed. Silicon and manganese do not also diffuse into the parent metal (Figure 2, *b*).

Investigations of wear resistance of the deposited metal were performed in machine NK-M [4] using the following conditions: abrasive – quartz sand with 0.2–0.4 mm sizes of particles; friction path – 400 m; pressure – 0.466 MPa; reference annealed steel 45. Figure 3 presents the averaged data about wear resistance of deposited metal obtained from results of three tests.

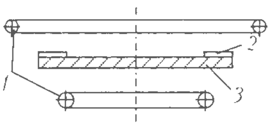
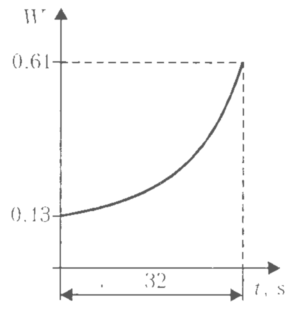
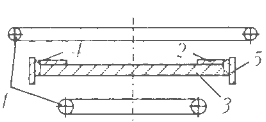
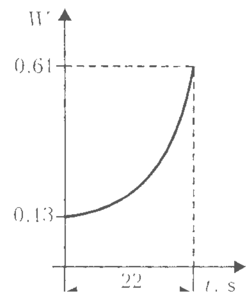
As is seen from Figure 3, the relative wear resistance and hardness of metal deposited by two technologies are almost at the same level.

Investigations of uniformity of thickness of deposited metal layer were made using procedure of [5]. Figure 4 presents the curve of a normal distribution of deposited metal layer across thickness. The non-hatched regions, located under the curve of a normal

Table 1. Microhardness (MPa) of structural constituents of deposited metal

Method of surfacing	*Chromium carbides*	*Matrix*	*White band (transition zone)*
Without screening	15320–17820	4550–5140	3220–3570
With screening	14300–15440	5150–5900	3780–4550

Table 2. Surfacing conditions with use of different heating systems

Method of surfacing	*Voltage, kV*		*Current, A*		*Surfacing duration, s*	*Variation of specific power at inductor W·10^9, W/m^3*
	in circuit	*at anode*	*tube mains*	*tube anode*		
Without screening of thermal and electromagnetic fields 	3.1–8.5	11.5–12.0	0.65–1.70	2.0–5.5	32	
With screening of thermal and electromagnetic fields 	3.1–8.5	11.5–12.0	0.65–1.70	2.0–5.5	22	

1 – inductor; *2* – hard alloy; *3* – component; *4, 5* – thermal and electromagnetic screens, respectively; generator VChI-63/0.44 was used in surfacing

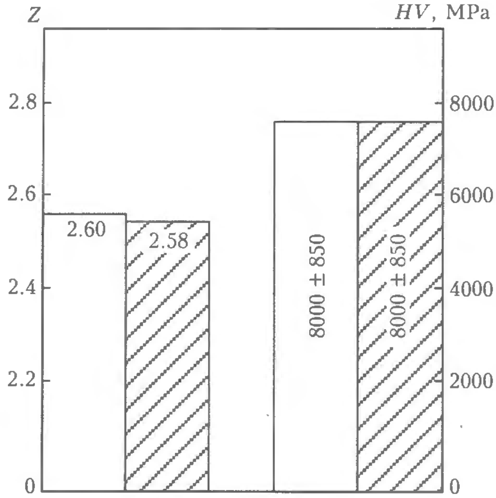


Figure 3. Relative wear resistance *Z* and hardness in metal samples obtained in induction surfacing without screening (*light columns*) and with screening (*hatched*) of thermal and electromagnetic fields

distribution, represent theoretically a percent share of components whose thickness of deposited metal is within the tolerance limit. Uniformity of distribution of metal layer across the thickness, deposited with screening of thermal and electromagnetic fields, is increased by 12 % as compared with surfacing without screening, owing to the more uniform distribution of power (temperature) in the surfacing zone. In addition, the electric power is saved by 10–15 % as a result of reduction in time of disc surfacing (from 32 to 22 s) and decrease in convective exchange between the edge surface of the component with environment (Table 2).

Thus, the results of investigations showed that the metal deposited by the induction method with screening of thermal and electromagnetic fields is not inferior by its properties to the metal deposited by a conventional technology. However, owing to screening the non-uniformity of distribution of deposited metal layer across the thickness is reduced by 12 %, additional saving of electric power is attained by 10–15 %, time of surfacing is reduced from 32 to 22 s,

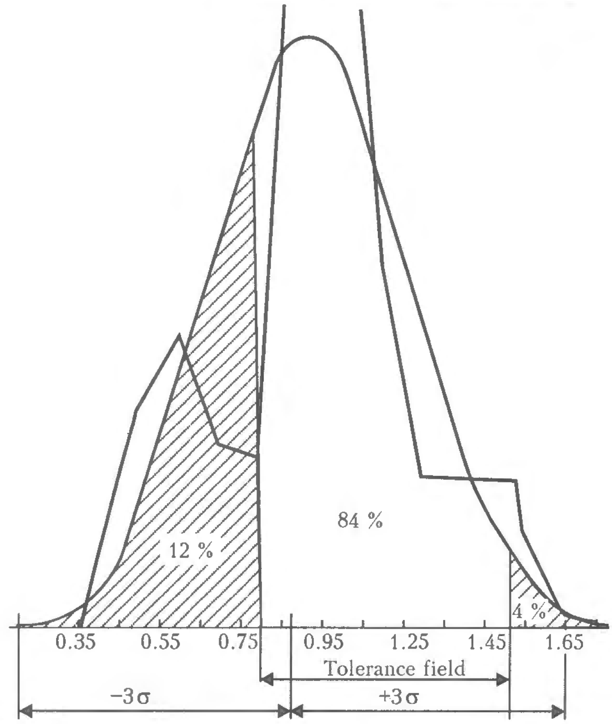


Figure 4. Curve of normal distribution of deposited metal layer obtained in use of heating system with screening

overheating of disc edge and deposited metal is eliminated.

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EFFECT OF SULPHUR ON PROPERTIES OF IRON-BASE ALLOYS AND PROSPECTS OF ITS APPLICATION IN SURFACING MATERIALS

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Literature data about the effect of sulphur on mechanical and service properties of steels and cast irons have been analyzed. It is shown that sulphur is a promising alloying element in surfacing consumables designed for repair and hardening of parts that operate under the conditions of friction of metal on metal at increased temperatures and are subjected to adhesion wear.

Keywords: *sulphur in steels, sulphur in cast iron, sulphide inclusions, mechanical properties of steels and cast irons, surfacing materials, alloying with sulphur.*

Sulphur is a harmful impurity in steels and other iron-base alloys. In particular, it forms fusible sulphide inclusions precipitating along the grain boundaries, thus leading to the appearance of crystalline cracks in metal, and causing also the hot brittleness of steel at temperatures of hot deformation [1]. Due to the negative effect of sulphur on the properties of steels and iron-base alloys its content is limited usually to 0.03 % [2].

From the data of works [3, 4] the sulphur is soluble unlimitedly in molten iron and possesses very low solubility in solid iron. As follows from Fe–S state diagram (Figure 1), the precipitation of the sulphide phase is not occurred in the process of an equilibrium crystallization of iron in the 1535–1365 °C temperature interval (δ-Fe) at sulphur concentration up to 0.18 % (that is its limiting solubility in δ-Fe at 1365 °C). The limited solubility of sulphur in γ-Fe at 1365 °C is 0.04–0.05 % and with decrease in temperature it is decreased and equals 0.005 % at 913 °C. $\gamma \rightarrow \alpha$-transformation at 913 °C leads to jump-like increase in sulphur concentration in iron up to 0.02 %,

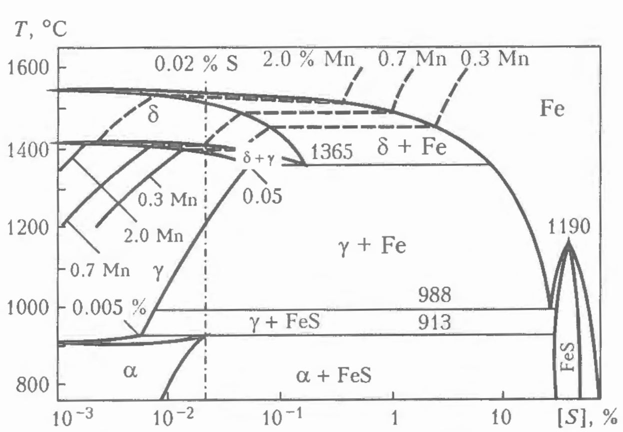

Figure 1. Fragment of diagram Fe–S (*solid curve*) and effect of manganese (*dashed curve*) on structural state [4]

however, its content in it at the further cooling is decreased again.

Sulphides, which are the main causes of negative effect of sulphur on the properties of steels and cast irons, are the compounds of sulphur with metals (mainly iron), and also with non-metals (boron, silicon, arsenic and others). According to work [5] the alloying elements can be arranged in the following sequence by susceptibility to the formation of sulphides in steel: Zr, Ti, Mn, Nb, V, Cr, Al, Mo, W, Fe, Ni, Co, Si.

Table 1 presents data of enthalpy of formation of sulphides of some elements by which different steels are alloyed [4].

In killed, well-deoxidized carbon steels, and also in alloyed steels the main mass of non-metallic inclusions are sulphides due to a lower concentration of oxygen in these steels as compared with sulphur concentration. Therefore, these are the sulphur inclusions that have a main influence on the steel properties, and not only composition of sulphide inclusions is important, but also shape [4]. In particular, from the point of view of reduction of probability of formation of crystalline cracks a globular shape of sulphides, uniformly arranged inside the grains is preferable.

By the composition, shape and arrangement in the steel structure, three types of sulphide inclusions can be distinguished [6, 7]: globular, chaotically arranged; inclusions of eutectic origin, forming films along the boundaries of grains; inclusions of crystalline shape, which are also chaotically arranged in the steel structure.

Work [8] describes the conditions of formation of different types of sulphides in carbon steel. The first type of inclusions is formed in a non-deoxidized steel (oxygen content of more than 0.02 %). Alongside with sulphur these inclusions can also contain oxygen, i.e. they are actually the oxysulphides. Sulphide inclusions of the second type are formed in steel, containing oxygen of less than 0.012 %, these are mainly the manganese sulphides.

Carbon and silicon increase greatly the activity of sulphur, decrease the temperature of liquidus and solidus and increase the interval of crystallization. In

Table 1. Enthalpies of formation of sulphides ($-H_{298}$, kJ/mol)

Alloying element	*Value of enthalpy depending on sulphide type*			*Alloying element*	*Value of enthalpy depending on sulphide type*		
	MeS	*MeS_2*	*Me_2S_3*		*MeS*	*MeS_2*	*$Me2S_3$*
Fe	95.46	185.06	–	Si	–	205.15	–
Co	85.41	140.26	213.53	Ti	217.71	334.94	–
Ni	92.95	142.35	–	Ce	494.04	644.35	1258.13
Al	–	–	723.90	Ba	443.80	–	–
Mn	205.16	207.25	–	Ca	469.55	–	–

low-carbon steels, deoxidized by weak deoxidizers, the sulphide inclusions of the first type are more often observed. With increase in carbon content by more than 0.15 %, the sulphides of the second type are appeared, while with increase by more than 1 % the sulphides of the third type are observed.

The general property of all three types of the sulphides is the fact that they are formed, as a rule, in crystallization of last portions of metal, enriched with liquation elements, both as those directly participating in the formation of sulphides (manganese, sulphur) and also those intensifying their activity (carbon, silicon).

Except the above-mentioned reduction in crack resistance and hot brittleness, sulphur can also influence negatively some strength characteristics of steels. The effect of sulphur content in the 0.020–0.135 % ranges on mechanical properties of cast carbon steel 35L of the following chemical composition was investigated, wt.%: 0.35C, 0.75Mn; 0.35Cr and 0.011P [4]. It was established that sulphur decreases greatly the ductility characteristics (elongation and reduction in area) and impact strength of steel 35L and does not almost influence the yield and tensile strengths (Figure 2).

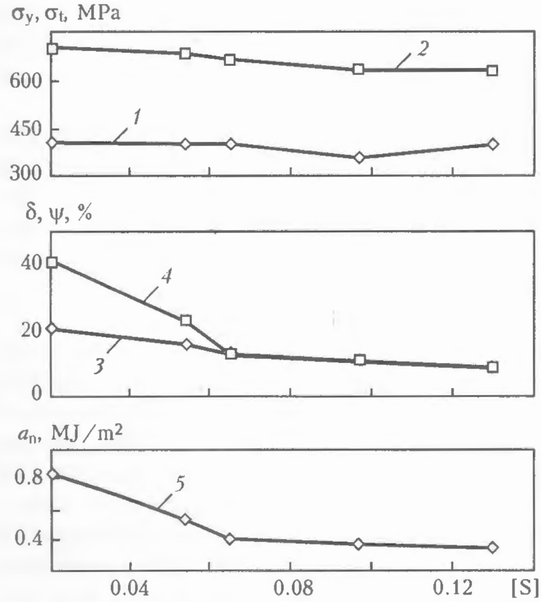

Figure 2. Effect of sulphur on mechanical properties of steel 35L (normalizing at 900 °C, tempering at 680 °C), deoxidation with 0.02 % Al (second type of inclusions) [4]: *1* — yield strength; *2* — tensile strength; *3* — elongation; *4* — reduction in area; *5* — impact strength (samples with a round notch)

It was also established that type of sulphide inclusions also influences the strength properties of steel 35L. Steel with the first type of inclusions has the maximum level of mechanical properties at all the contents of sulphur and the minimum level is typical of steel with the second type of inclusions. Moreover, the significant reduction in ductility and impact strength is observed with increase in sulphur content up to 0.05 %, its further increase has a lower influence on the above-mentioned parameters.

Works [9, 10] show the dependence of impact strength of steel 16GS on content of carbon, manganese and sulphur. Sulphur shows the highest effect on the impact strength. At its content of more than 0.03 % the impact strength is decreased by 30 % (Figure 3). Data of works [9, 10] confirm also that yield and tensile strengths of steel 16GS are not changed with increase in sulphur content.

Investigations of different authors [4, 11–15] confirmed that sulphur has also the similar effect on mechanical properties of cast low- and high-alloy steels. In addition, sulphur reduces the characteristics cold brittleness of steel, the critical temperature of cold

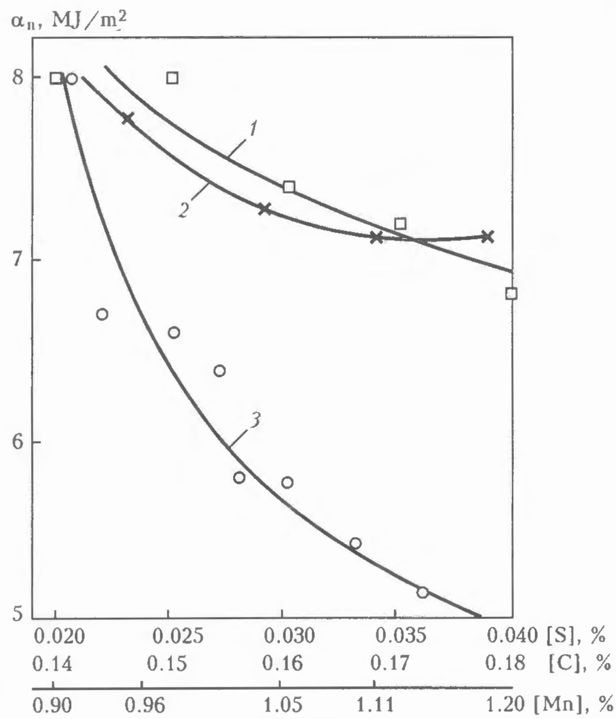

Figure 3. Dependence of impact strength of steel 16GS on content of carbon (*1*), manganese (*2*) and sulphur (*3*) [10]

Table 2. Chemical composition (wt.%) and mechanical properties of chromium steels with increased sulphur content [15]

Type of steel	*C*	*Cr*	*S*	σ_t, *MPa*	σ_y, *MPa*	δ, %	ψ, %	a_n, MJ/m^2
A	0.12	12.1	0.02	494	222	31.8	73.7	2.41
B	0.12	12.0	0.23	531	237	32.6	59.2	2.36
C	0.12	12.1	0.51	501	239	28.5	58.4	1.49

brittleness of cast steel is shifted to the side of positive temperatures with increase in sulphur content [4]. Similarly the sulphur influences the mechanical properties of steel after rolling. This is proved by data of [15] on example of steel of Kh12 type (Table 2). Steel with an increased content of sulphur has the lower characteristics of ductility and impact strength. These characteristics are especially low in samples cut out across the rolled metal.

There is a large number of publications about the effect of sulphur on the properties of welds and deposited metal [16–24]. It was outlined in works [16, 17] that sulphur is one of main causes of initiation of hot (crystalline) cracks in welding. Depending on alloying the formation of cracks is considered to be due in the first turn to the formation of eutectics Fe–FeS (temperature of solidification is 988 °C), Ni–NiS (temperature of solidification is 645 °C) or others, having also low temperature of melting [17, 22 and others].

To reduce the probability of formation of crystalline cracks, it is recommended to use steels and welding consumables with a decreased content of sulphur for welded structures. At the same time the feasibility of prevention of hot cracks by alloying the welds of steels of Kh18N9 and Kh25N20 type with a large amount of sulphur (up to 3 %) is considered in the work [17]. This sulphur content makes it possible to avoid the crystalline cracks in welds owing to the effect of «curing» (filling) of intercrystalline gaps with excessive fusible sulphide eutectics.

The carried out analysis showed that sulphur has mainly the negative effect on many characteristics of steels and alloys on iron base, and also welds and deposited metal, however, there are also data about its positive effect on some properties of steels. Thus, the sulphur alloying is used for improving the workability owing to the decrease in adhesion of tool to the metal being subjected to working [3, 15, 25].

There is information in work [15] about stainless steel, containing 12 % Cr and 0.4 % S, which is used for manufacture of products of mass production, where a good workability in automatic lathes and the presence of a smooth surface are required. It is considered that the inclusions of sulphides contribute to the better removal of chips in cutting, thus improving the surface of the workpiece treated.

Data are known [26, 27] about positive effect of sulphur on wear-resistance of grey cast iron. It was established [26] that when the grey iron is alloyed with sulphur up to 0.86 %, its wear resistance is many times increased and by this characteristic it becomes to be superior to high-strength and malleable cast irons. Authors of the work recommend the sulphurous grey cast iron for manufacture of machine parts operating under severe conditions of friction and wear with lubricant and without it.

Similar data were obtained in work [27] on grey iron, alloyed by 5 % Cu and 0.5 and 1 % S. The investigations showed that this cast iron possesses high wear-resistant properties under the conditions of dry friction and can be a substitute for high-alloy cast irons.

The data are known about application of sulphur in welding and surfacing consumables [25, 28]. In works [25, 28] the effect of 0.0250–0.329 % S on structure and properties of deposited high-speed steel, close by composition to steel R9M4K8, was studied. Metal cutting tool, deposited by high-speed steel, containing 0.32 % S, had 1.5–2 times higher wear resistance than a serially-manufactured tool. The productivity of treatment was increased almost by the same times.

From some data [29] the sulphides in metal friction on metal can play a role of a solid lubricant, thus improving the tribotechnical characteristics of the friction pair. In addition, they prevent the adhesion of metals, in particular at increased temperatures. Thus, sulphur can become promising as an alloying element for some types of hardfacing materials. The most acceptable field of application of these materials can become the hardfacing of cutting and die tools, mill rolls of some types and other parts operating under the conditions of metal to metal friction and subjected to adhesion wear.

The negative effect of sulphur on the properties of deposited metal can be neutralized to a certain extent by transfer of sulphides from film to a globular shape, i.e. from the second type to the first type. These sulphides should be arranged uniformly in the alloy matrix.

It should be noted in the conclusion:

• sulphur is used for alloying the steel and iron-base alloys to improve the workability and quality of surface of parts. However, with increase in its content, the characteristics of impact strength, ductility and cold brittleness of steel are decreased, but the yield and tensile strengths are not changed in this case;

• sulphur in steel and iron-base alloys forms fusible eutectics precipitating along the grain boundaries, which are one of causes of crystalline crack initiation. At high content of fusible eutectics the effect of «curing» of formed tears is observed;

• sulphides prevent the adhesion wear, in particular at increased temperatures, moreover, they can play a role of a solid lubricant, thus improving tribotechnical characteristics of the friction pair. Thus, sulphur can be challenging for alloying of materials in hardfacing of dies and die tools, mill rolls of some types and other parts operating under the metal to metal friction at increased temperatures and subjected to adhesion wear.

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FEATURES OF THE PROCESS OF ESH WITH A COMPOSITE ROD IN A SMALL-SIZED SECTIONED MOULD

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Electrophysical and thermal peculiarities of electroslag hardfacing (ESH), using an additional hollow graphite electrode, are considered. It is shown that addition of a nonconsumable electrode to the slag pool leads to formation of a high-temperature region in it, which allows a uniform melting of different components of hardfacing consumables. Use of ESH enables producing a heat-resistant deposited alloy based on nickel aluminide.

Keywords: *electroslag hardfacing, sectioned mould, composite rod, simulation of ESH process, hollow graphite electrode, current distribution, heat balance, nickel aluminide*

Electroslag hardfacing in a sectioned mould (SM) of small-sized (up to 50 mm diameter) end faces of parts of equipment and tools allows producing high-quality metal as a result of effective metallurgical processing of the melt by overheated slag and ensuring its directional solidification at a high enough efficiency of hardfacing with various materials without applying current to them. Particularly effective is ESH in SM for strengthening of broaching tools, operating under the conditions of cyclic temperature-force impact at up to 1100 °C temperature [1]. However, under such complex conditions, use of many modern nickel and cobalt superalloys is not effective enough [2].

Known are new types of alloys, produced by various metallurgical processes, characterized by higher technological and service properties at operating temperatures, for instance, with a matrix based on titanium and nickel aluminides. In individual cases such alloys found application in industry [3]. It is known [2] that high-temperature composite alloys based on Ni_3Al nickel aluminide are more readily adaptable to fabrication and usually contain a number of alloying elements with different melting temperature: 660 (aluminium), 1453 (nickel), 3410 (tungsten), 3000 (tantalum), 2610 °C (molybdenum). In order to produce deposited metal of such a type with a high level of welding technological properties, it is necessary to use composite hardfacing materials, consisting of a metal shell and filler, in which the high- and low-melting components are contained in the form of metal powders and wires. This provides a stoichiometric relationship of alloying elements in the deposited metal. Such a material can be a composite rod, consisting of a nickel shell and filler. However, effectiveness of using a composite rod in the regular ESH process is low because of a non-uniform melting of its components in the slag.

Depending on the circuit of applying current to slag (through a consumable metal and nonconsumable graphite electrode or current-conducting section of the mould) ESH processes induce a diverse and non-uniform thermal condition of the slag pool melt. In consumable-electrode ESH, in particular with flux-cored wire [4], maximum temperature (about 2300 °C) of the slag is observed in a small local zone (heat center), adjacent to the electrode, which is where it melts. In the case of a nonconsumable electrode it is difficult to precisely feed the filler material into the heat center of the slag. In ESH in SM the heat center is in the peripheral section of the slag pool at the walls of the current-carrying section of the mould [5], the other part of the slag being heated up to the temperature of 1900 °C [1], which is sufficient for a uniform melting of the composite rod components.

In order to ensure a sound melting of the composite ingot, studied was the possibility of formation of a new heat center in the slag pool by adding another hollow nonconsumable electrode to it.

Conditions of running of the ESH process were simulated in water-cooled SM and a one-piece mould for the cases of different current supply to the slag. Slag pool was simulated by electrolyte solution, where the viscosity (0.03 Pa·s) approximately corresponded to that of overheated slag (ANF-6 flux). Material for the models of SM and one-piece mould were three copper rings of 50 mm diameter and 2.5 mm thickness, which were placed into a glass cavity of a cylindrical shape (Figure 1). Current-carrying section ring had a vertical cut. In order to simulate the one-piece mould, the rings were connected to each other, and in SM model they were insulated. A copper cylinder of 20 mm diameter and 7 mm height was used as the workpiece. Graphite electrodes of 5, 12 and 15 mm diameters with a solid and hollow section were used.

In all the experiments, a voltage of 36 V from DC current source was applied to the current carrying section of the model mould, and 28 V voltage from an independent power source was applied to the additional graphite electrode. After immersion of the graphite electrode into the electric bath (electrolyte) to different depths, also its motion in the volume was registered visually. In order to observe the nature of

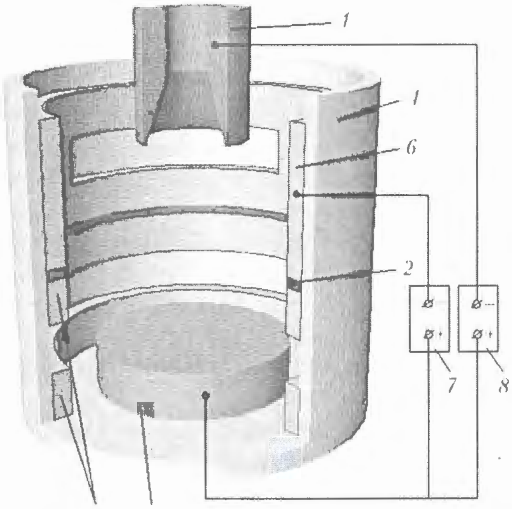

Figure 1. Model of SM with an additional electrode: *1* — hollow graphite electrode; *2* — insulator; *3* — item; *4* — glass cavity; *5* — forming section; *6* — current-carrying section; *7, 8* — DC power source

electrolyte motion, it was dyed with graphite chips. In view of the complexity of recording the electrolyte flows in the model in the small volume of the pool, following its motion is difficult, but possible at the initial moment of the model functioning (for 8 to 10 s). Direction of the vector of electromagnetic force $\mathbf{f}_e$ was determined for a sound evaluation of the direction of slag motion, taking into account the data of [6].

In order to study the forecast high-temperature region in the slag, it is necessary to evaluate the current distribution in the slag pool, which influences the mode of slag flow and determines the thermal pattern in it. Distribution of current lines in the model schematics was established proceeding from the data of works [5, 7, 8], obtained for different cases of current supply to the mould. Studied model schematics allows determining the possibility of creating a high-temperature region in SM slag pool by adding a hollow nonconsumable electrode into it.

Experiments conducted under the actual conditions showed that during ESH the working end face of the hollow graphite electrode takes the shape of an «inverse» cone (Figure 2). In this case, the area of the active surface of the electrode is reduced, this providing a higher density of current. Investigation results indicated that addition of a nonconsumable electrode to SM significantly changes the energy situation in the slag pool (Figure 3). Electric field of the current-carrying section of the mould ousts the current lines from the nonconsumable hollow electrode into the underelectrode region, where their high concentration is created. Such a distribution of current increases slag heating in the zone of electrode immersion, thus inducing a temperature gradient in the slag pool. A high density of the current lines on the internal surface of the conical electrode tip provides heating of slag in this region up to the maximum temperature, thus promoting equalizing of the rate of melting of the composite rod components.

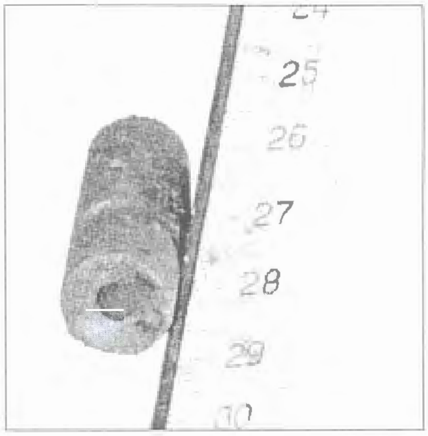

Figure 2. Appearance of the working end face of a hollow graphite electrode

The high density of current in the lower part of the wall in the current-carrying section and at the electrode surface allows reaching here the maximum value of the bulk electromagnetic force, directed downwards and causing vortex motion of the electrolyte. It may be assumed that the model of electrolyte motion will have the form of two flows, directed towards each other (see Figure 3). A smoothing effect from electrolyte rotation in the horizontal plane is observed in the small volume of the model. Such a rotation is due to the impact of the magnetic flux on the electrolyte as a result of the change of current direction in the split current-carrying section. In this connection, the actual pattern of the slag flow can be distorted, and its motion can only be considered in the instantaneous variant. Flow of electrolyte *10* meeting condition rot $\mathbf{f}_e > 0$, while moving along the electrode and further on to the metal pool, gives away part of its heat, heats the pool and rises to the slag pool center. Here, flows *10* and *11* take the same direction, which creates a resultant flow on their boundary, which entraps and carries the metal drops

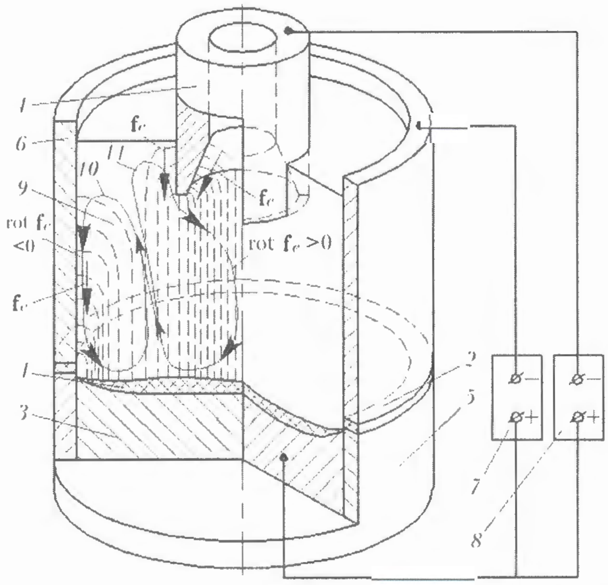

Figure 3. Schematic of current distribution and slag motion in the studied (Figure 1) model of SM: *1, 2* and *5–8* — see designations in Figure 1; *3* — deposited metal; *4, 9* — metal and slag pools, respectively; *10, 11* — slag flows (dashed lines show current lines)

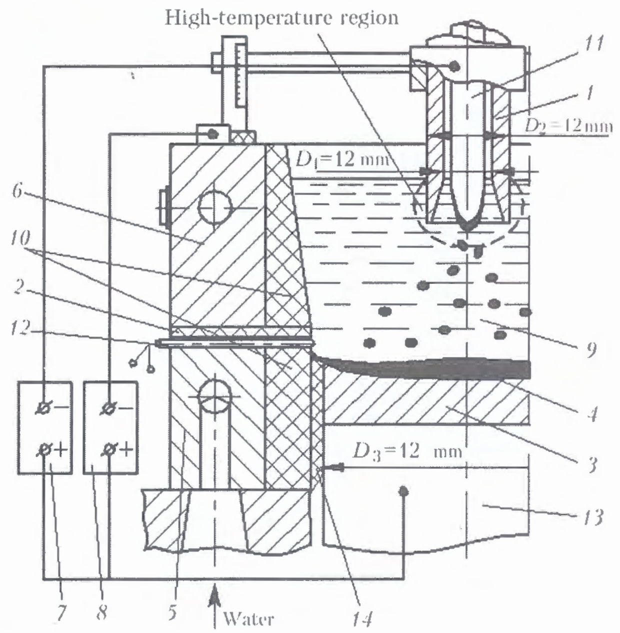

Figure 4. Schematic of ESH in SM with a hollow graphite electrode: *1–9* — see designations in Figure 3; *10* — graphite insert; *11* — composite rod; *12* — thermocouple; *13* — item; *14* — slag lining; D_1, D_2 — inner and outer diameter of the hollow graphite electrode, respectively; D_3 — deposited item diameter

to the mould walls. Moving along the SM near-wall region, flow *11* cooled in its upper part, for which rot $\mathbf{f}_e < 0$, is heated in the lower part of the wall of the current-carrying section, while transferring the heat to the metal pool surface. Such a motion results in longer duration of the metal drops staying in the slag, and longer time of their exposure to the slag.

Experimental hardfacing was performed by ESH schematic, developed as a result of modeling (Figure 4). Used as a hardfacing material was a rod of 5 mm diameter, consisting of the nickel sheath, which was filled with powders of aluminium, zirconium, molybdenum boride, graphite, and wires of commercial-grade tungsten, tantalum, molybdenum and Np-Kh20N80T grade. Addition of wires of refractory components eliminates separation of the light and heavy metal powders in the charge. Composition of the composite rod was calculated, proceeding from the composition of the known alloy, containing nickel and aluminium in the stoichiometric proportion, providing

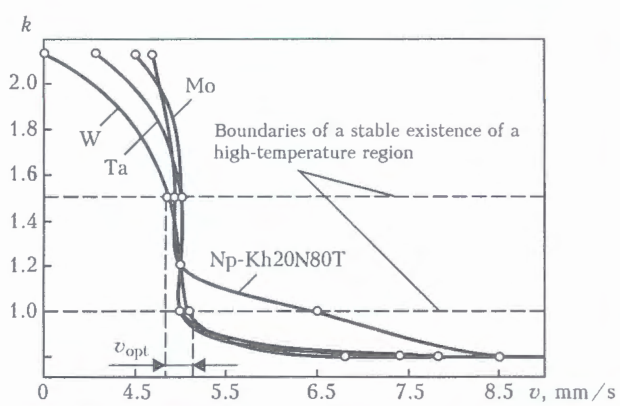

Figure 5. Influence of the ratio of currents *k* on the electrodes on rate *v* of the composite rod melting

nickel aluminide Ni_3Al. Other alloying elements in the alloy were taken in the following proportions, wt.%: 2.5–3.0 W; 0.5–2.5 Ta; 2.5–3.0 Mo; 0.5–2.5 Zr; 4.4 Cr; 0.035 B. Graphite powder was added, allowing for the amount of carbon, going into the slag as a result of electrochemical dissolution of the working part of the graphite electrode. Weight of metal deposited on 40Kh steel billet was 40 g. Rate of melting of each wire component in the composite rod was evaluated by the amount of metal, melted per a unit of time. Optimum mode of ESH of the composite rod was that, in which the difference in the rate of its component melting is minimum.

ESH was conducted with DCSP (VDU-504 and VDU-1000 power sources), using ANF-6 flux. During hardfacing the current from the current-carrying SM section I_{SM} and current from the hollow graphite electrode I_e was varied in the range of 140–250 A, and slag voltage in the range of 17–23 V. Height of the slag pool was maintained in the range of 25–30 mm. The process was conducted in argon. Slag temperature was recorded using tungsten-rhenium thermocouples of VR 10/20 grade with a multichannel potentiometer KSP-4. Hardness of the deposited metal at normal and high temperatures was determined in TSh-2 instrument with a hard-alloy sphere of 5 mm diameter at 7.35 kN load and application for 10 s. Metallographic studies of the deposited metal were conducted with the known methods.

Investigations demonstrated that the ratio of currents $I_{SM}/I_e = k$ has the strongest influence on the uniformity of melting of the components in the composite rod and stability of ESH process. At $k \leq 1.5$ the slag temperature in the near-electrode region rises up to 3300 °C. In ESH the slag starts boiling at the electrode surface, which, however, does no disturb process stability. Further increase of *k* values leads to intensive boiling of the slag pool and its splashing, which is negative for running of ESH process. At $k \leq 1$ the slag temperature in the near-electrode region drops and becomes comparable with that of the slag pool in SM. Therefore, during hardfacing *k* ratio ensures a stable existence of the high-temperature region in the zone of the electrode immersion, and should be equal to 1.0–1.5. This range of *k* values is valid for I_{SM} and I_e currents equal to 140–250 A. If current values go beyond the above range, ESH process becomes unstable. If the above value of *k* ratio was maintained, the optimum rate v_{opt} of composite rod melting was equal to 4.8–5.2 mm/s (Figure 5), this allowing sound deposited metal to be produced.

Heat balance was calculated in order to compare the thermal mode of SM operation by the traditional and studied sequences. It was assumed that the heat pattern in the slag is determined by the action of two independent sources of its heating. A different temperature of the slag in the zone of the graphite electrode immersion and in the rest of the slag pool was assumed in calculations. Analysis of the obtained data (Figure 6) showed that in consideration of the ESH

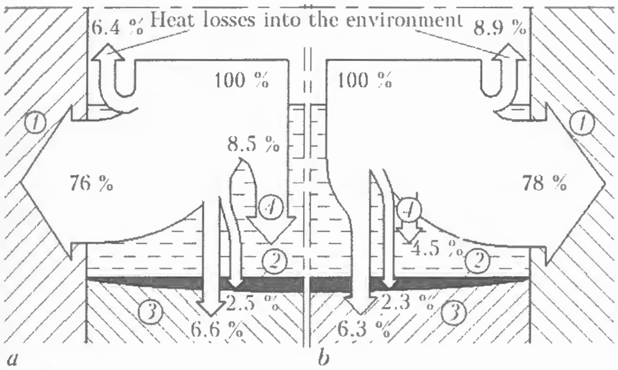

Figure 6. Diagram of heat balance of ESH process in SM with an additional hollow nonconsumable electrode (*a*) and by the traditional schematic (*b*): *1–4* — distribution of heat flows in the mould, metal pool, billet and composite rod, respectively

process schematic the heat consumption for heating and melting of the base metal is close to that in the traditional schematic, and almost 2 times more heat is consumed for heating and melting of the composite rod. This is due to the high temperature gradient in the slag in the zone of electrode immersion, which results in faster heating of the surfacing material, thus allowing a more effective consumption of the slag pool heat. Considerable and similar heat losses to the mould in both the cases are related to the slag pool contacting it over 70 % of its surface.

Metallographic investigations demonstrated that the deposited metal does not contain any welding defects and has a complex heterophase structure, the basis of which is a solid solution based on nickel aluminide Ni_3Al. The deposited metal structure contains dispersed particles based on the solid solution of γ-phase nickel, which contains the secondary γ′-phase, which is Ni_3Al. Located along the dendrite boundaries are eutectic γ′-phases, carbides and carboborides, the composition of which includes nickel, aluminium and other alloying elements. Total content of Ni_3Al based phases is equal to about 80 % (Figure 7, *a*). Composite structure of the deposited metal with directional arrangement of the dendrites provides the required thermal fatigue and hardness (*HB* 100–110) at up to 1100 °C temperature, which allows using it under the conditions of cyclic temperature-force impact.

The zone of deposited metal fusion does not have any defects, or critical solidification and diffusion interlayers, affecting the service properties of the hardfaced item (Figure 7, *b*).

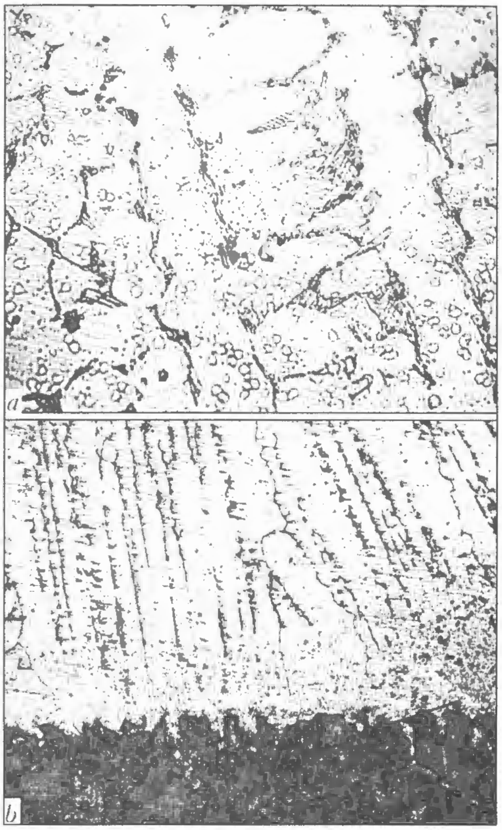

Figure 7. Microstructure of deposited metal (*a* — ×400) and zone of fusion with the base (*b* — ×100)

CONCLUSIONS

1. At addition of a hollow graphite electrode into the slag pool of the small-sized SM, a high-temperature region forms in the zone of electrode immersion into the slag, the stability of this zone being provided by 1.0–1.5 ratio of currents of current-carrying section of the mound and hollow graphite electrode.

2. Developed schematic of ESH in SM with an additional hollow graphite electrode allows obtaining a sound high-temperature resistant deposited metal based on Ni_3Al for operation under cyclic temperature-force impact conditions at temperature of up to 1100 °C.

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COMPUTER SYSTEMS OF INFORMATION SUPPORT OF WELDING FABRICATION

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Described are the information support systems, developed by the E.O. Paton Electric Welding Institute for welding and surfacing technologies.

Keywords: *welding and surfacing technology, welding consumables, welding modes, welding aerosols, metallurgical equipment, certification of specialists, information systems, expert systems*

The diversity of the methods of welding, consumables, types of welded joints, groove shapes and spatial position of the weld, base metal thickness and composition, variants of the modes and other conditions of welding, make it necessary to develop modern systems of information support of welding fabrication, designed for specialists working in the Chief Welder units in engineering plants. Alongside the information-retrieval, document, manager and other information systems, of special interest is development of computer systems of welding technology design, which would accumulate the available knowledge, data and gained experience of highly-qualified experts in the welding field. It is rational to perform development of such systems so as to successively «cover» individual subject areas of welding fabrication with information systems.

Let us briefly dwell on analysis of the existing approaches to development of information support systems. The first form of computerized ordering of information were databases, organized in the form of files of machine-readable data. Despite the evolution over the past 40 years, all the databases are limited by inclusion of just passive information, replying to the question of *what*. The passive nature of the databases consists in the following: when replying to the question of *what*, these information systems imply that their user knows the answer to the question of *how*, i.e. knows how these data could be used to solve the posed problem. If active information answering the question of *how*, is entered into the computer memory in addition to passive information, the database is thereby complemented by the knowledge base. The thus plotted information systems are called expert systems. Specialists of the British Computer Society give the following definition of an expert system: «An expert system means a systems combining the computer capabilities with knowledge and experience of an expert in such a form, that a system may suggest an intelligent solution of the posed problem». Such a rather general and broad definition covers the most diverse computer systems, however, in any case, in development of an expert system, presence of two figures is important, namely expert on the subject and knowledge engineer. The first of them has the function to provide raw information in the form of data and knowledge, required to develop an expert system in a particular problem area. Knowledge engineer systematizes, processes and generalizes the obtained information, develops models of databases and knowledge bases, designs and develops the respective software.

The question of *how*, to which expert systems reply, has two extensions: *what to do* — so-called expert systems, pertaining to factual expertise; *how to interpret* — cognitive expert systems, pertaining to interpretation expertise. Factual expertise ensures solving the engineering problems, while interpretation expertise is a tool meant for scientific workers and aimed at designing new technological processes.

Development of a specific expert system always gives rise to two questions: 1) how complete and valid is the information which the expert has on the databases and knowledge bases in terms of the possibility of solving the posed problem, ensuring the validity and quality of computer solutions; 2) what should we do with the fact, that part of the information inevitably becomes obsolete, and how can we keep the system in an updated condition as long as possible. When replying to the first question, the expert's experience and qualifications are of paramount importance. At the same time, however, the most successful solution of the problem of raw information validity is achieved by development of quantitative models of databases, which allow screening the certainly invalid information on a formal level. The problem of maintaining the information system in an updated condition is solved to a certain extent by establishing special database editors, which enable the user to independently supplement the bases with new data, correct the available data or eliminate the data which are not urgent for a specific user, configuring the information system in the necessary direction.

Ideology of expert systems turns out to be extremely attractive for informatization of welding fabrication. This paper describes some information systems in the field of welding, developed over the last years at the E.O. Paton Electric Welding Institute.

Expert system SURFACING [1] (expert is I.A. Ryabtsev). The system is designed for development of technologies of mechanized electric arc surfacing for 12 surfacing processes for parts of the type of bodies of revolution and those with a flat surface, operating in different industries (in metallurgy, mining industry, chemical engineering, agriculture, etc.). Structurally the system consists of several subsystems, namely selection of surfacing material; development of surfacing technology; editor of databases of surfacing materials; and bank of surfacing technologies. The subsystem of surfacing material selection uses the database of surfacing materials, which in the author's version contains information on 280 materials used in CIS countries. The latter includes the data on application of a surfacing material (operating conditions and kinds of material wear), welding-technological characteristics of the material, wear resistance of the deposited metal, etc. Parts operating under similar conditions, are classified in 42 groups (Figure 1), so that it is enough for the user to indicate the group, which includes the part to be surfaced, so as to get the recommendations on suitable surfacing materials. For a more accurate determination of the optimum surfacing material, it is necessary to additionally enter into the computer detailed information on the operating conditions and kinds of part wear. In case there is an alternative for material selection the system ensures information support for the user in the form of data (Figure 2) on the composition, wear resistance and welding-technological properties of deposited metal. Such a support enables the user, proceeding from individual quality criteria made of surfacing of this part, selecting a suitable surfacing material. The database of surfacing materials is kept in an updated condition, using a special editor, allowing the user to independently enter the data on the new surfacing materials or edit the existing information.

The subsystem of designing the surfacing technology gives recommendations on suitable processes and technique of surfacing this part and depending on the part type, its overall dimensions and spatial position of the deposited layer, the system suggests to the user variants of rational modes of surfacing with welding wires (strips) of different diameter. The final result of operation of SURFACING expert system is the output document, which shows the results of joint work of the welding technologist and computer on development of the surfacing technology. The data bank of technical solutions allows accumulating in the computer memory recommendations on the technology of surfacing the individual parts and components, obtained using the SURFACING expert system, as well as correcting this information by the results of test verification, thus creating in the factory a bank of typical surfacing technologies.

Expert system for development of the technologies of surfacing parts of metallurgical equipment [2] (expert is I.A. Ryabtsev). The system is child product of SURFACING expert system aimed at designing the technologies of surfacing parts of metallurgical equipment. The databases and knowledge bases of the system contain information on surfacing several hundred parts of metallurgical equipment. In-

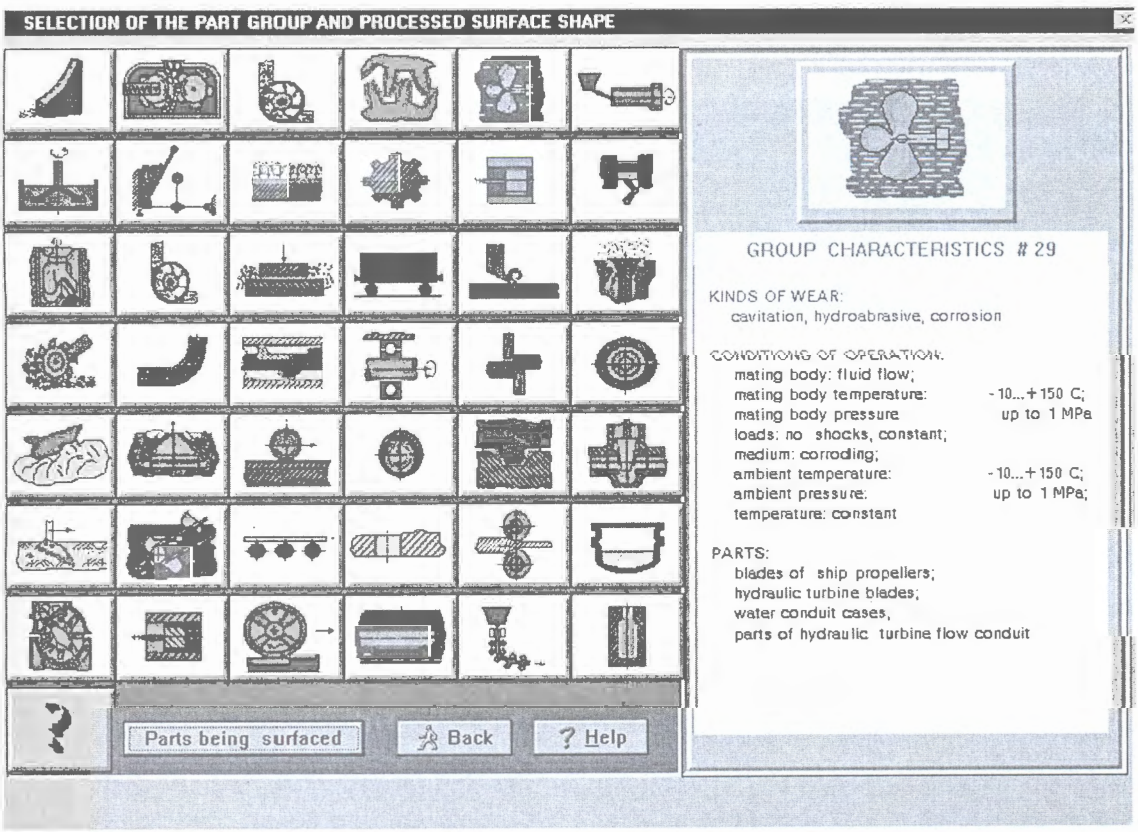

Figure 1. Classification of surfaced parts by groups

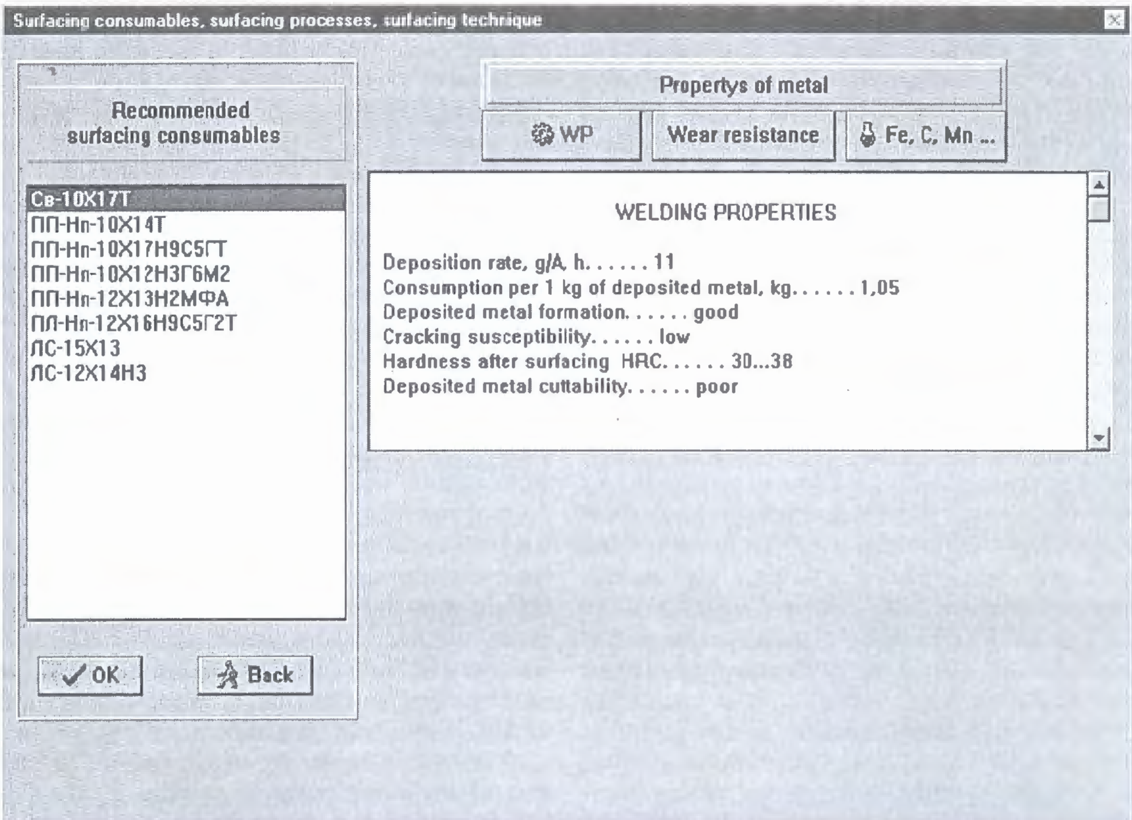

Figure 2. Information support of solution on selection of surfacing material

formation retrieval is organized by a hierarchy sequence «Production» (shop of metallurgical enterprise)» — «Installation» — «Part» (Figure 3).

Expert system for designing technologies of light alloy welding (experts are A.Ya. Ishchenko, V.P. Budnik). Developed expert system accumulated the knowledge, practical experience and data on technologies of welding light alloys implemented in the form of databases and knowledge bases. The expert system includes the following factographic and graphic databases: base materials; welded joints; welding consumables; welding processes and modes; comparative characteristics of welding processes. Database of welded joints includes information on weld

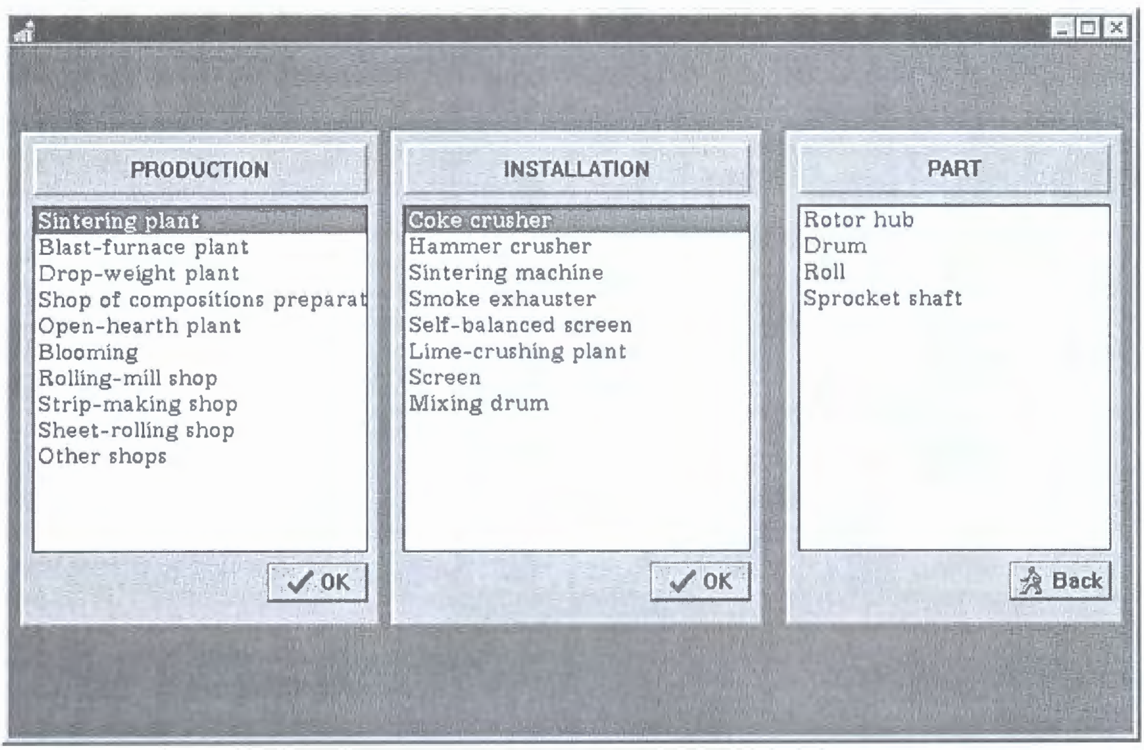

Figure 3. Selection of part to be surfaced

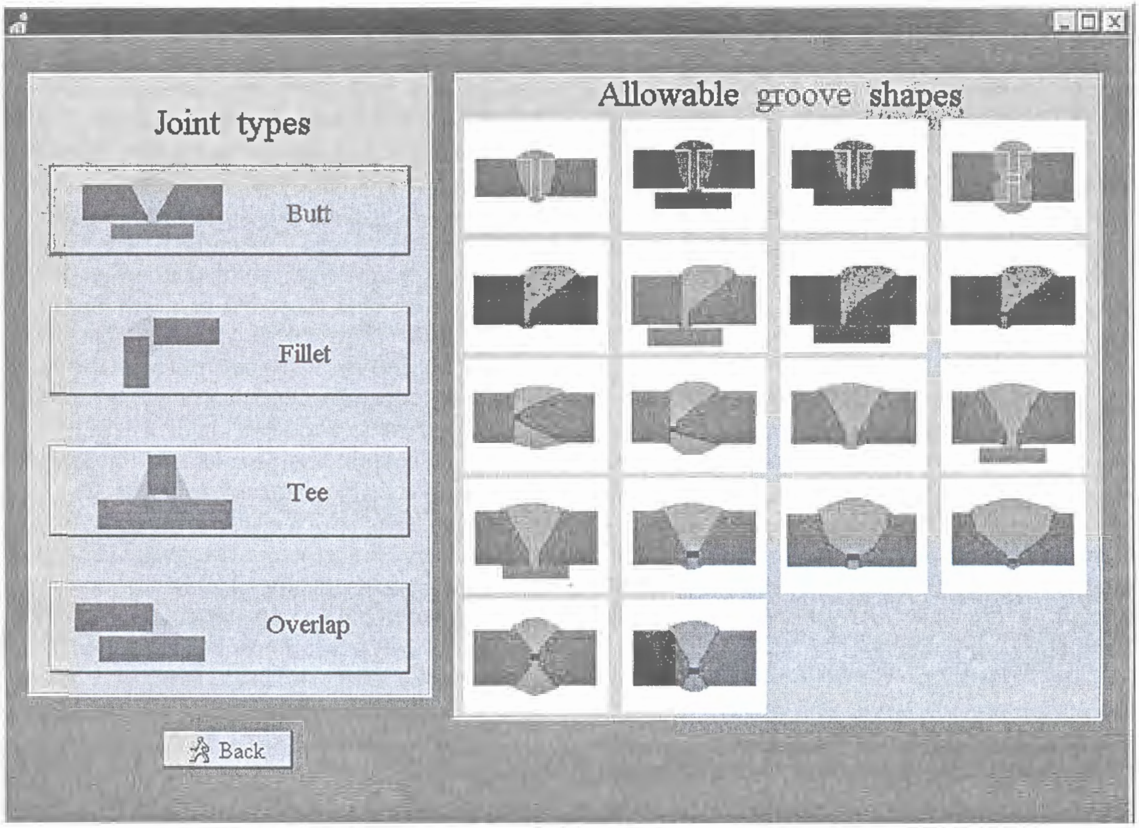

Figure 4. Selection of joint type and groove shape in light alloy welding

groove shapes for butt, fillet, tee and overlap joints, admissible processes of welding with a particular groove shape, as well as geometrical limitations for metal thickness and weld length. As an illustration of the system operation, Figure 4 shows an element of user dialogue with the computer, during which the groove shape is chosen for a butt joint with the specified base metal thickness. Expert system allows selecting welding consumables, process and modes of welding for different alloying systems and base metal grades.

Data bank of hygienic characteristics of welding aerosols [3] (experts are O.G. Levchenko, V.A. Metlitsky). Information on welding aerosols is concentrated in the databases composed for coated electrodes, welding wires and fluxes. They contain information on factors, influencing the composition and level of welding aerosols (filler material grade, wire diameter and type, welding process and mode, composition of a shielding gas mixture, etc.), as well as data on levels of evolution of gaseous and solid components of welding aerosols. In order to form databases and regularly update them during operation of the computer system, a database editor was developed, which allows performing their initial filling and subsequent editing.

Information retrieval in the databases is performed by the specified grade of the welding consumable, the result is produced as an output document, which presents the data on welding conditions and hygienic parameters, characterizing the composition, level of precipitation and toxicity of welding aerosols. Capacity of general ventilation by dilution is calculated by the data on limit admissible concentration of welding aerosols. Recommendations are issued on equipment for work place ventilation and means of individual protection of welder's respiratory organs. Owing to a detailed description of welding conditions and level of evolution of welding aerosols by the results of computer system operation it is possible to perform hygienic assessment of the welding process for each grade of welding consumable, issue specific recommendations on application of the respective means of welder protection and perform complete analysis of the influence of all the factors on the quantitative and qualitative composition of welding aerosol, as well as its toxicity. The latter enables conducting hygienic optimization of the process and technology of welding, optimizing the welding consumable composition at the stages of their development, selecting less toxic of the available welding consumable grades.

Data bank of the modes of CO_2, submerged-arc and inert gas welding (expert is P. Zeyffarth). The data bank accumulates information on the welding modes, depending on base metal thickness, welding process and position, welded joint type. For instance, for semi-automatic CO_2 welding the data on the modes cover up to 200 mm thicknesses with information on the number of passes, welding modes for the root, filling and decorative welds, and with X-shaped groove these data are given for filling the lower and upper groove. In a number of cases, the welding modes are illustrated by weld macrosection, this enabling the user selecting the most rational from his viewpoint mode from alternative variants. The database currently contains 1200 entries of welding modes.

Computer system for expert certification (experts are P. Zeyffarth, A. Sharf). The system is a certain unified shell, designed for certification and training of participants in training courses. The system includes three modules: module of forming examination questions, question cards and teaching material (appropriate database editor); module of testing trainee knowledge with teaching elements; examination module performing official certification of trainee knowledge by individual sections of the course. The system supports several training courses, each of which has a certain number of sections. Examination question contains variants of answers, one or several of which are correct. Four different formats of replies to examination questions are envisaged: textual replies and graphic variants of the replies, including those with a figure, clarifying the question. The editor enables forming the bases of examination questions for diverse training courses. The lecturer generates question card from them, selecting at his discretion questions pertaining to various sections of the course. During operation of the examination module, the computer through the random number sensor selects a card from the card set, to which the trainee is to reply during the assigned time. Correctness of the replies is evaluated by a 100 point scale. Test module is used for preparing the trainees for taking the examinations, and it can also be used as an auxiliary means during the training process. Unlike the examination, at testing the question card is generated by the random number sensor from the examination question base. In this case, the trainee is able not only to control the correctness of his/her replies, but also get information support in the case, if the reply turned out to be incorrect. For this purpose, the editor of the database of examination questions correlates the question and the respective section of electronic manual, which the lecturer prepares in advance, using special software. In this case, there is also the capability of direct access to the electronic manual through its contents. The developed computer system can be used in the chairs of higher educational establishments, industrial enterprises, secondary education establishments for certification of trainees and specialists. In particular, it is intended to use it in training and certification of specialists in the course of European welding engineers. With this purpose the experts prepared the bases of examination questions for several training courses, and developed a special electronic manual. Assessing the prospects of development and use of electronic means of information support of welding fabrication, the authors believe that the computer systems described in this paper, alongside other systems, already developed [4] or to be developed in the future, will allow creating in engineering plants a computerized work place, thus permitting the Chief Welder units using the experience and knowledge of highly qualified experts.

Acknowledgement. *The authors are sincerely grateful to experts for many years of fruitful co-operation in development of the above systems of information support in the field of welding.*

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PLASMA-POWDER CLADDING OF WEAR- AND CORROSION-RESISTANT ALLOYS IN VALVE MANUFACTURING

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Results of investigation of alloys used in valve manufacturing for cladding of stop valve components are given. Peculiarities of plasma-powder cladding are considered. It is shown that plasma-powder cladding provides the best performance of seal surfaces.

Keywords: *plasma-powder cladding, corrosion-resistant alloys, valves, seal surface, structure, hardness, penetration*

As shown by operating experience, failure of different-application stop valves is caused primarily by damages of their seal surfaces. Scores, cracks, erosion and corrosion fractures [1] may be formed on them, thus breaking tightness of a stop valve.

Cladding using special alloys is a reliable method to ensure high performance of seal surfaces. Service conditions of the valves are very important for selection of the alloys. Metal deposited on the seal surfaces should be resistant to scores in its sliding friction on metal in different working environments under a specific pressure of up to 150 MPa. The deposited metal should have a minimal sensitivity to cracking in thermal cycling with a temperature difference of up to 200–300 °C. Hardness at working temperatures should be sufficiently high (not less than *HV* 300). The deposited metal should be not inferior in erosion and corrosion resistance to stainless steel of the 18-8 type. Its structure and properties in long-time operation should be stable, and it should feature good machinability and high technological effectiveness. Only very few of wear- and heat-resistant alloys meet these requirements. Cobalt-base alloys with chromium and tungsten additions are most suitable for cladding the seal surfaces of the valves. Nickel-base alloys with chromium, silicon and boron additions can also be used for some types of the valves.

Co–Cr–W alloys — Stellites. These alloys were developed by E. Haynes (USA) in 1907–1908, and were first described by him in 1913. More than 60 grades of Stellites are available now. They are characterised by high erosion and corrosion resistance, can retain their high hardness to a temperature of 650 °C and have good wear resistance [2–5].

Structure of deposited metal, where the content of chromium is normally 30 wt.%, depends mostly upon the content of carbon and tungsten. Hypoeutectic Co–Cr–W alloys are utilised for cladding the valve components. Structure of these alloys consists of solid solution and carbide eutectic containing carbides of the M_7C_3 and $M_{23}C_6$ types. Ageing (5000 h, 700 °C) results in partial decomposition of carbides M_7C_3 and precipitation of dispersed carbide $M_{23}C_6$ in solid solution, this leading to increase in hardness.

Alloys of the following chemical composition are used most extensively for cladding, wt.%: 1.0–1.6C, 1.5–2.5Si, 28–32Cr, 4–8W, ≤ 3Fe, Co — base. Hardness of the deposited metal is *HV* 400–480. Harder alloys with a higher content of carbon and tungsten are sensitive to cracking at high thermal cycling. The alloys have satisfactory machinability in cutting and grinding. Because of a high sensitivity of Co–Cr–W alloys to hot cracking, a part to be clad should be preheated to a temperature of 600–750 °C, and this temperature should be maintained during cladding. The clad parts must be subjected to annealing at 800–850 °C for 2–3 h and then slow cooling. These peculiarities, as well as a high price of cobalt, add to the cost of the clad parts. Drawbacks of Co–Cr–W alloys include also an insufficiently high score resistance [3], which limits specific working pressure on the valves. In addition, it is undesirable to use the Co-containing alloys within the radiation zones, as cobalt forms long-lived radioactive isotope ^{60}Co.

For plasma cladding of valve components, the E.O. Paton Electric Welding Institute developed an alloy of the Co–Cr–W–C system additionally alloyed with boron, which forms a low-melting point boride eutectic capable of healing hot cracks. Its positive effect is well-known from the experience of welding heat-resistant austenitic steels and alloys. Alloying of Stellites with boron decreases their melting point, facilitates production of powders used as additives and improves formation of the deposited metal. Also, boron effectively increases hardness of the deposited metal at temperatures of up to 550–600 °C (Figure 1) and score resistance. Tests to heat resistance in air at 800 °C for 1000 h and in water vapour (700 °C, 0.1 MPa, 500 h) revealed no substantial difference in the character of oxidation of Stellites without and with boron (up to 2.6 wt.%). In distilled water (300 °C, 20 MPa, 1000 h), the alloys without and with boron (2.6 wt.%) have the rate of corrosion equal to 0.0038 and 0.0028 g/(m²·h), respectively.

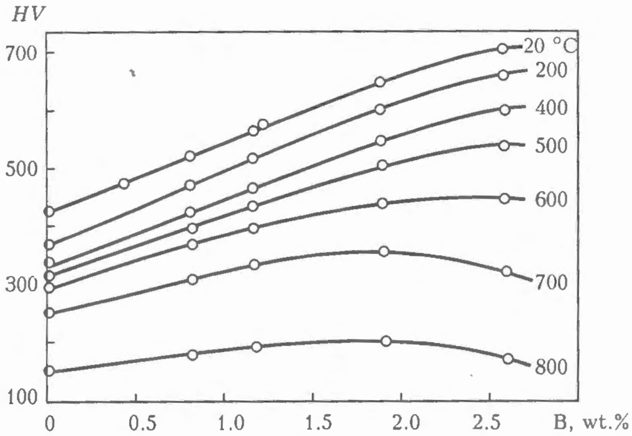


Figure 1. Dependence of hardness of deposited Stellite upon the content of boron at different test temperatures

Investigations of the effect of boron on crack resistance of Stellite 6 showed that boron content of the deposited metal should be not in excess of 1.2–1.5 wt.%. It is recommended that cladding of the seal surface of the valves should be performed using powder PN-AN34 (PrKKh30V5N6SR) of the following chemical composition, wt.%: 0.7–1.0C, 1.5–2.5Si, 28–32Cr, 4–5W, 4–8Ni, 0.5–0.9B, ≤ 3Fe, Co — base. Hardness of the deposited metal is *HRC* 44–48. This powder allows cladding of the valve components with a diameter of up to 150 mm without preheating. For larger components, preheating to no more than 500 °C is sufficient.

Nickel alloys containing chromium, silicon and boron. More than 90 alloys of this type are available now. Part of them has a known trademark name «Colmonoy». These alloys emerged in the 1950s and gained acceptance as self-fluxing brazing filler alloys and filler metals for gas cladding, as well as flame metallising followed by fusion. This type of the deposited metal has a good wear resistance, retains high hardness to a temperature of 600 °C, and is resistant to corrosion in many aggressive environments [6, 7]. Other advantages of Ni–Cr–Si–B alloys include comparatively low melting point (1050–1150 °C), moderate cost compared with Co–Cr–W alloys, and possibility of substantially decreasing preheating temperature in cladding. These alloys are successfully used for cladding some types of the valves [1, 6].

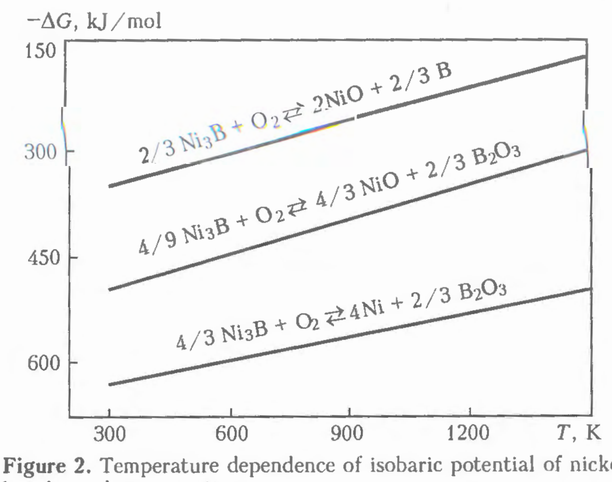


Figure 2. Temperature dependence of isobaric potential of nickel boride oxidation reactions

Structure of a C-free alloy consists of solid solution, boride eutectic with microhardness *HV* 415–445 and borides Ni_3B, CrB and Cr_5B_3. The quantity of eutectic grows with increase in the concentration of carbon, and primary hexagonal carbides M_7C_3 are formed at a carbon content of more than 1 wt.%. Carbon is also part of carboborides. Hexagonal crystals with microhardness *HV* 1040–1385, H-shaped carboborides with microhardness *HV* 1775–2340 and crystals of irregular shape with microhardness *HV* 2700–3800 are detected in structure of the alloys.

All chromium is contained in Ni-base solid solution. Increase in the concentration of chromium leads to formation of chromium borides CrB in the alloy. Chromium and boron enter into reaction in such a proportion that the composition of solid solution remains almost unchanged (16–18 wt.% Cr). Silicon is almost completely contained in solid solution and forms no individual phases. It increases reactivity of chromium in solid solution, thus favouring the formation of chromium carbides and borides. Almost all boron contained in the alloys is in the form of borides. Iron is mostly included into the composition of solid solution. Increase in the iron content leads to decrease in the amount of eutectic, which is indicative of increase in the solubility of carbon and boron in this solution.

Self-fluxing of alloys of the Ni–Cr–Si–B system is especially important for cladding of parts with preheating, where a surface to be clad is covered with an oxide film. Boric anhydride and silica, while entering into reaction with the substrate metal oxides, dissolve them, thus creating conditions for good wetting of the deposited surface. In addition, boron and silicon are favourable for excellent formation of the deposited layer. Ni-base alloys, containing (wt.%) 0.35–0.70C, 2.5–3.5Si, 10–14Cr and 2B, are however inferior in resistance to corrosion in water vapours at a temperature above 500 °C to cobalt Stellites.

Resistance of alloys of the Ni–Cr–Si–B system in water vapours is found to substantially depend upon the chemical composition, and is determined by the character of borides in structure of an alloy. Nickel boride Ni_3B is formed in the alloys at a low chromium content (less than 14 wt.%). As reported by Omori and Hashimoto [8], calculations of the Gibbs energy of possible reactions of oxidation of Ni_3B indicate that the most probable reaction is that with formation of Ni and B_2O_3 (Figure 2). The melting point of B_2O_3 is 468 °C. At the presence of a water vapour, it forms a well-soluble and volatile orthoboric acid H_3BO_2. This results in loosening of the oxide film, thus leading to corrosion acceleration.

Examination of steam valve components, the seal surfaces of which are clad with alloys of the PG-SR3 type, which were in use in a water vapour atmosphere at a temperature of 545 °C and pressure of 25.5 MPa for 7000–31216 h, showed that the mean rate of corrosion of metal was 0.06 mm/year, and the formed oxide layer had a loose structure.

Alloy with an increased chromium content (more than 18 wt.%), in which chromium boride CrB is formed, behaves in a different way. Oxidation of this alloy leads to formation of borate glass Cr_2O_3–B_2O_3, which is refractory and resistant to water. Increase in the content of chromium is accompanied by increase in its concentration in solid solution, which also provides a substantial increase in heat resistance of the alloy. The investigations conducted resulted in the development of Ni-base alloy N68Kh21S5R of the following chemical composition, wt.%: 0.35–0.50C, 4–5Si, 20–22Cr, 1.0–1.3B, 4–7Fe, Ni — balance. The alloy has low sensitivity to scores, good thermal stability and high corrosion resistance [9].

Plasma-powder cladding is the most appropriate method for cladding the valve plates and other components with a good access to the surface treated [1, 5, 10]. Owing to mechanisation of the process, plasma-powder cladding provides a higher and more consistent quality of the deposited metal, compared with manual acetylene-oxygen and electric-arc cladding. Advantages of this method include considerable increase in productivity, saving of cladding consumables, improvement in labour conditions and reduction of post-cladding machining. One-layer cladding is advisable in terms of achieving the maximal productivity of the process and saving of cladding consumables. Here the share of the base metal should be kept at minimum, if the iron content of the deposited metal is restricted, as is the case of cladding of the Co–Cr–W system alloys.

The productivity of electric arc cladding measured by the amount of metal deposited per unit time under the set technological conditions is unambiguously determined by the value of the electric current. Increase in the current is accompanied by increase in the productivity and, at the same time, increase in the share of base metal in the deposited layer. In practice this leads to the necessity to either limit the current or use multi-layer cladding. In both cases the time of cladding a workpiece increases. Consumable electrode is melted mostly due to the heat released in the active arc spot. The electrode metal is transferred to the weld pool in the form of drops overheated to a temperature much higher than the melting point. For example, for electrode wire or rod of low-carbon steel the temperature of the drops is 2300 °C, and it grows to the iron boiling point with increase in the welding current.

In plasma cladding, the powder additive particles in the form of grains get into the arc column in the cold state and are heated there due to heat exchange with plasma. Temperature of the particles at the moment when they touch the weld pool or workpiece surface depends upon their shape and size, thermal-physical properties of metal, time of their dwelling in plasma and parameters of the latter (temperature, velocity and thermal conductivity). Grains of a powder may get into the weld pool either in solid or molten form, depending upon the above factors. If the mean-mass temperature of the powder is lower than the mean temperature of the weld pool metal, increase in the powder feed speed will decrease penetration of the base metal.

Based on the theory of thermal processes of welding by N.N. Rykalin, it can be shown that the share of base metal in the deposited layer will be low (≤ 5 %) if the relationship between the effective thermal power of the arc, q, amount of a powder additive, G_{ad}, fed to the weld pool and its heat content S_{ad} meets the following condition:

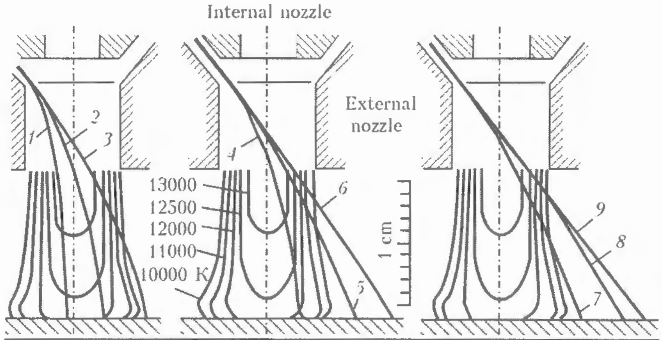
$$G_{ad} = k \frac{q\eta_t}{S_{melt.ad} - S_{ad}(1 + \eta_t)},$$

where η_t is the thermal efficiency of the process of penetration by the arc without a powder; $S_{melt.ad}$ is the heat content of the metal additive at a melting point, including the latent melting heat; and k is the correction factor allowing for deviation of the calculation diagram from actual conditions (k = 1.2–1.5).

It follows from the equation that, with maintaining a shallow penetration of base metal, the productivity of plasma cladding can be increased either by raising arc power q or by increasing the efficiency of heating of a powder in the arc. The powder is heated in the arc primarily as a result of convective heat exchange, the intensity of which depends upon many factors, such as plasma parameters, shape and size of the particles, etc. The time of heating is determined by the path and velocity of flight of the particles, as well as by the size of the arc column.

Figure 3 shows the temperature field of plasma and the calculated paths of powder particles having different diameters: 0.10, 0.25 and 0.50 mm, at a particle material density of 8.1 g/cm^3. Operating parameters of the plasmatron are as follows: non-transferred arc current — 80 A, transferred arc current — 150 A, plasma gas (argon) flow rate — 2 l/min, transporting gas (argon) flow rate — 9 l/min, and axial velocity of plasma flows at the nozzle exit section — 160 m/s. Small-diameter particles can be readily accelerated in the arc, their paths bend and pass through the high-temperature zone of the arc. Coarse particles change but slightly their velocity and flight paths under the effect of the plasma flow. The time of their dwelling in the arc depends first of all upon the angle of introduction of a powder to the arc and initial velocity.

Under the above conditions of operation of the plasmatron the temperature of the particles, depending upon their size and initial velocity at the moment when they touch the surface treated, ranges from 300–500 °C to the melting point of an alloy (Figure 4), according to the calculations. Increase in the temperature of plasma enhances heating of the powder. According to spectroscopic measurements in a region where the powder is heated, the temperature of plasma is determined primarily by the transferred-arc current



Calculated path of particles	1	2	3	4	5	6	7	8	9
Initial velocity, m/s	2.0	4.6	8.0	2.0	4.6	8.0	2.0	4.6	8.0
Particle diameter, mm	0.10			0.25			0.50		

Figure 3. Paths of flight of powder particles in plasma arc

(Figure 5). Therefore, increase in the current of this arc leads to increase not only in the heating of a workpiece, but also in the efficiency of heating of the powder. This provides fast growth of the productivity of plasma-powder cladding with increase in the electric current.

Relationship between the current, share of base metal in the deposited layer and productivity of cladding alloys of the Ni–Cr–Si–B system on low-alloy steel is shown in Figure 6, *a*. Similar data were obtained also in cladding alloys of the Co–Cr–W system on austenitic Cr–Ni steel (Figure 6, *b*) [11]. As the productivity grows, the range of the current ensuring permissible penetration of base metal widens (Figure 7). Preheating of the base metal required to avoid hot cracks in the deposited metal during plasma cladding does not prevent ensuring a shallow penetration (Figure 8). This circumstance is especially important for cladding alloys of the Co–Cr–W system, as the permissible iron content of the deposited metal is restricted to avoid dramatic deterioration of its properties.

Quality of the powder used as additive is very important for plasma-powder cladding. The powder should have good flowability, required particle size composition and low gas saturation. Shape of the particles, spherical or close, is optimal for the powders used for plasma-powder cladding, as it provides the maximal flowability of the powder. Powders used for plasma cladding have a particle diameter of d = 80–200 μm in the most. Using a coarser powder leads to increase in its losses and deterioration of quality of

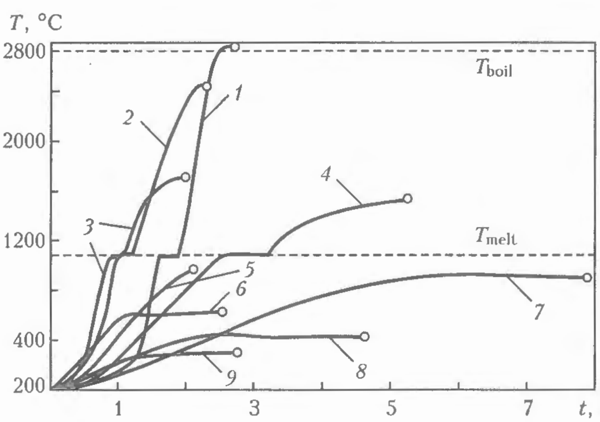


Figure 4. Heating of Ni–Cr–Si–B alloy powder in plasma arc: *1–9* — see Figure 3; the moment of contact of a particle and surface treated is marked by circles

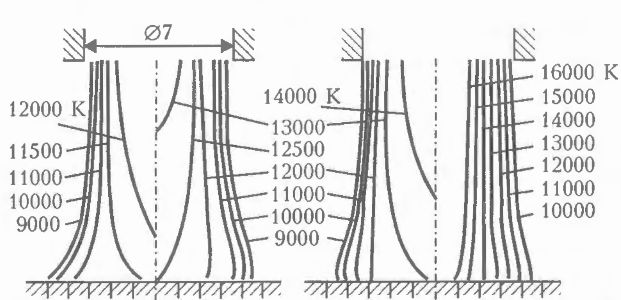


Figure 5. Temperature field in argon plasma arc column at different values of electric current

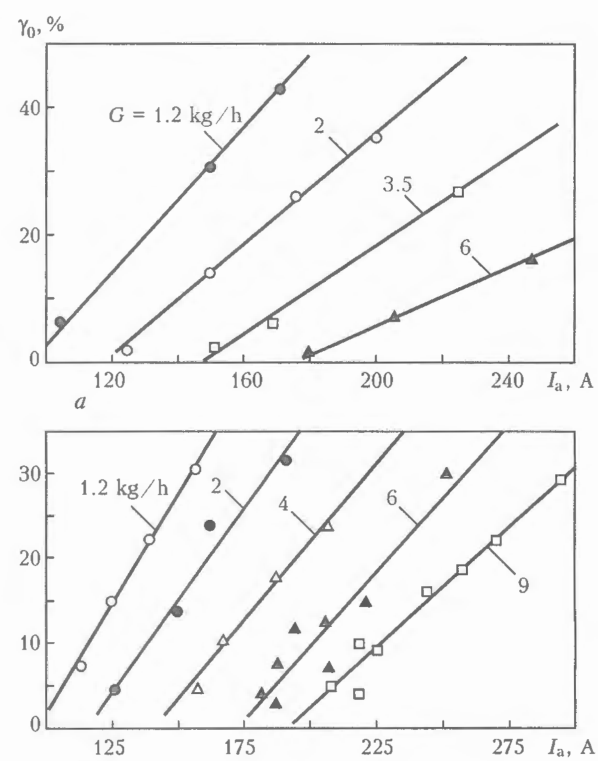


Figure 6. Dependence of the share of base metal, γ_0, in deposited layer upon arc current I_a at different deposition efficiencies using powders of the Ni–Cr–Si–B (*a*) and Co–Cr–W (*b*) systems

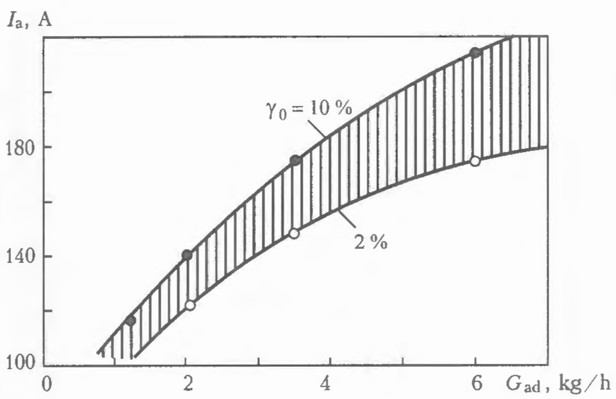


Figure 7. Range of current where the share of base metal in the deposited metal is 2–10 wt.% (cladding of alloys of the Ni–Cr–Si–B system on low-carbon steel)

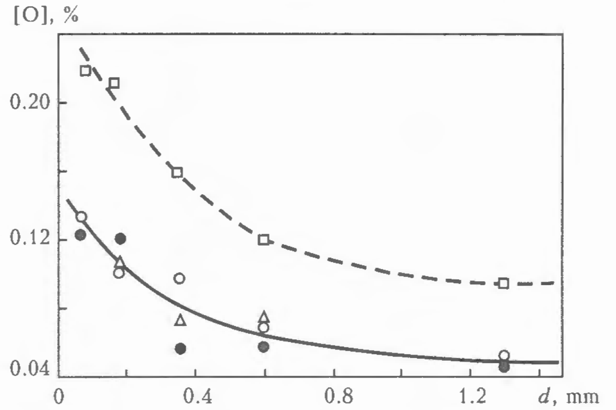


Figure 9. Oxygen content of particles of the Ni–Cr–Si–B alloy powder produced by nitrogen (○), argon (Δ), water (●) (*solid curve*) and air (*dashed curve*) atomisation

the deposited layer. If a powder is too fine, it melts inside the plasmatron, thus leading to clogging of its nozzles and violation of the cladding process.

Powder additives contain gases dissolved in metal, combined into chemical compounds (oxides), which are absorbed on the surface or fixed inside hollow particles. An increased content of gas, especially oxygen, in a powder causes spattering of metal during cladding, formation of pores and non-metallic inclusions in the deposited layer, and formation of slag on the bead surface. As shown by the investigations, powders of the Ni–Cr–Si–B system intended for plasma cladding should have an oxygen content of not more than 0.10–0.12 wt.%, and in powders of the Co–Cr–W system this content should not exceed 0.006-0.800 wt.%. The above requirements are best met by the powders produced by atomisation of liquid metal in inert gases (Figure 9).

Plasma cladding is performed using a combined-type plasmatron (Figure 10). Heating of a powder and workpiece is done by the plasma transferred arc. A non-transferred arc performs auxiliary functions. The powder additive is fed with gas via a flexible tube from a hopper to the plasmatron, and is blown into the arc through an annular slot between the stabilising and focusing nozzles. Three flows of a working gas are fed to the plasmatron: central plasma gas flow, which protects the tungsten electrode from oxidation, stabilises and constricts the arc (the gas flow rate is 1.5–2.0 l/min); transporting gas flow, which feeds the powder additive (the gas flow rate is 4–6 l/min); and shielding gas flow (the gas flow rate is 8–12 l/min). The working gas is argon.

Technological capabilities of plasma cladding using powder additives are very wide. Minimal thickness of the layer in plasma cladding is 1 mm, and maximal — 5–6 mm (one-layer bead). To deposit wide beads, transverse oscillations are transmitted to the plasmatron (bead width is up to 60 mm).

The E.O. Paton Electric Welding Institute developed several types of units for mechanised plasma cladding of valve components. The versatile unit OB2184 (Figure 11) is intended for cladding of cylindrical surfaces with a diameter of up to 400 mm and up to 800 mm long, and flat parts 800 × 500 × 400 mm in size. Cladding can be done also on conical and shaped parts. The unit is completed with device A1756, which can be used independently.

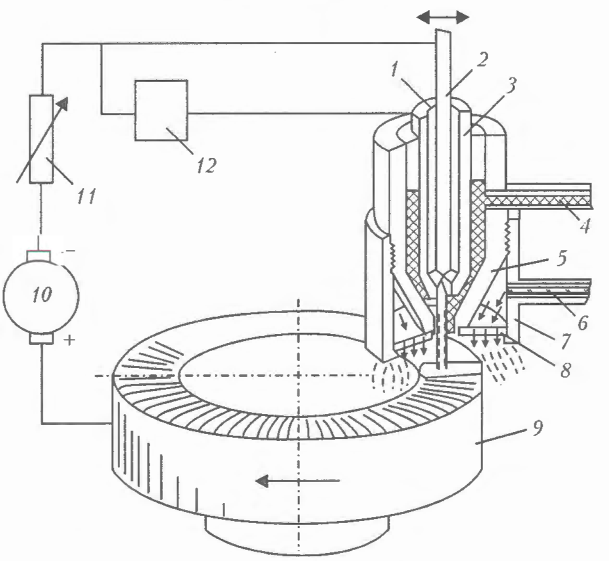


Figure 10. Schematic of the combined-type plasmatron: *1* — introduction of plasma gas; *2* — tungsten electrode; *3* — stabilising nozzle; *4* — introduction of transporting gas with powder additive; *5* — focusing nozzle; *6* — introduction of shielding gas; *7* — shielding nozzle; *8* — gas lens; *9* — workpiece; *10* — arc power supply; *11* — resistor; *12* — arc ignition device

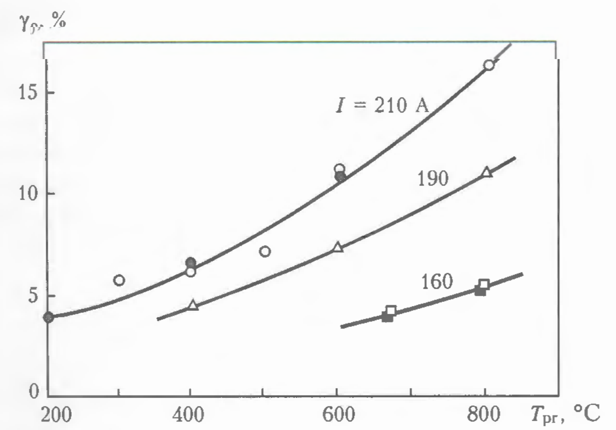


Figure 8. Dependence of the share of base metal, γ_0, upon preheating temperature T_{pr} at different currents (cladding of the Co–Cr–W system alloy on steel Kh18N10T, productivity — 6 kg/h)

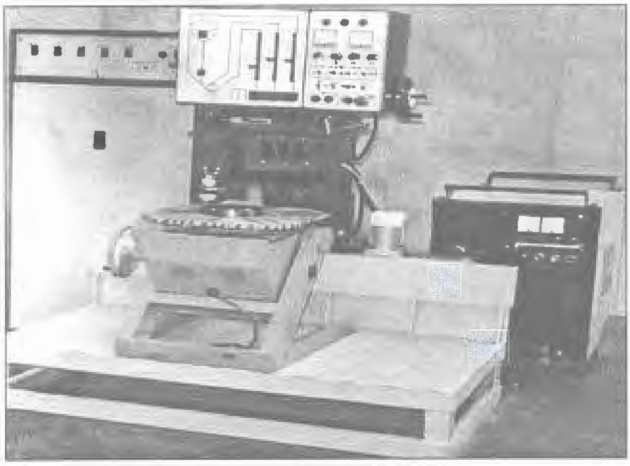


Figure 11. Versatile unit OB2184 for plasma-powder cladding

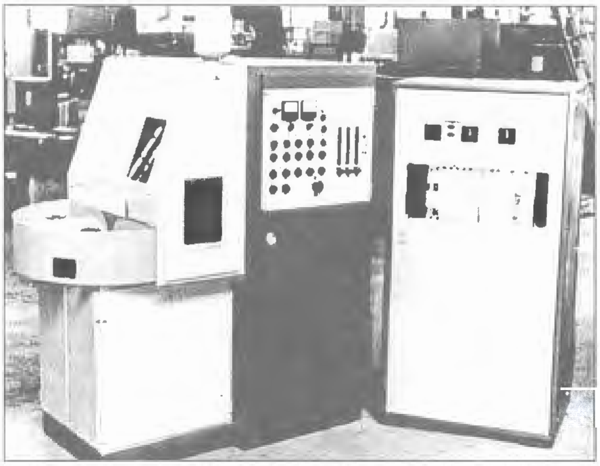


Figure 12. Automatic unit UP142 for cladding of valve components

It can be easily mounted on a bracket, which allows cladding on valve components with D_n 1000 mm or more.

Automatic unit UP142 (Figure 12) can serve as an example of the specialised unit, intended for cladding of the disk-type parts with a diameter of up to 200 mm and up to 150 mm high. The unit is controlled from microprocessor, which is necessary to ensure the automatic cladding cycle. It is fitted with a work chamber that provides reliable protection of an operator from light and thermal radiation of the arc, gas and dust. The unit is also equipped with a mechanism for automatic unloading of billets after cladding to a collecting bin. It can be readily integrated with the line of automatic loading and unloading of parts in mass production. The control electric circuit of the unit has an option of performing cladding in one or two layers.

The many-year experience shows that plasma cladding provides the high quality and homogeneity of the deposited metal on seal surfaces of valves (Table 1).

Hardness was measured at five points uniformly distributed in width of the beads, with a spacing of 4–5°. Statistical processing of the resulting data was performed in an assumption that the distribution of hardness followed the normal law. The higher the variation coefficient that represents the ratio of standard deviation to mean value of hardness, the higher the non-uniformity of the distribution. Manual cladding provides the least uniform deposited metal. In submerged-arc one-layer cladding using alloying flux the hardness is uniform in width of the bead, but its distribution on the perimeter of a workpiece is highly non-uniform. This is attributable to heating of a workpiece during the cladding process and increase in the

Table 1. Results of measuring hardness of deposited metal on stop valve components (D_n 150–175 mm)

Cladding method, additive material	*Quantity of measurements*	*Mathematical expectation of hardness, HRC*	*Standard deviation of hardness, HRC*	*Variation coefficient, %*
Manual three-layer cladding with electrodes TsN-6M	431	30.8	3.82	12.3
Manual three-layer cladding with electrodes TsN-12M	496	39.3	4.34	11.0
Submerged-arc wire cladding using alloying flux	351	29.8	2.30	7.8
Plasma-powder cladding using powder Pr-N73Kh16S3R3	417	48.6	1.28	2.62

Table 2. Characteristics of reliability of treated components

Type of deposited metal	*Failure rate, 1/ths h*	*Mean life to failure, ths h*
180K62Kh30V5S2 (electrode TsN-2)	0.245	4.8
	0.099	10.1
Kh16N7G4S4M5B (electrode TsN-12)	0.094	10.6
	0.049	20.8
N73Kh16S3R3 (powder)	0.02140	46.7
	0.00475	210.0

Note. Data for plate are given in numerator, and data for seat are given in denominator.

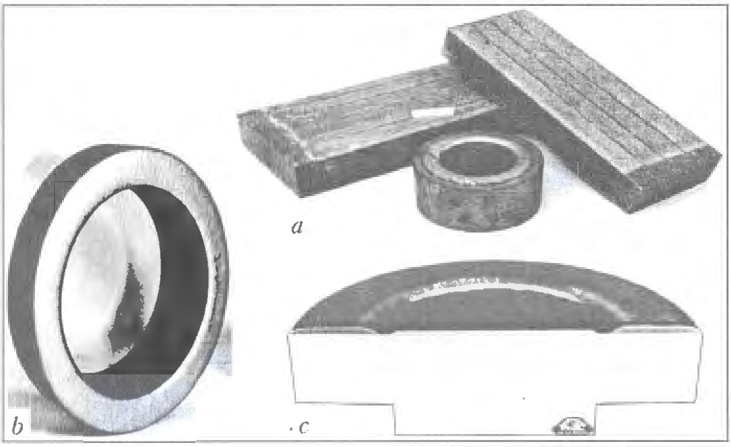


Figure 13. Clad components of Christmas trees (*a*), clad seat (*b*), and macrosection of clad plate (*c*) of the power generation valve gates with D_n 150 mm

depth of penetration of base metal, as well as to the presence of the fading out regions, where cladding is performed actually in two layers. In addition, alloying of the deposited metal through flux is sensitive to variations in cladding parameters, which is inevitable under industrial conditions.

The efficiency of using different alloys for cladding seal surfaces of power generation valves can be estimated most objectively using quantitative reliability criteria. Because of unavailability of express methods for comprehensive reliability tests of the valves, the use is made of the only source of information on their reliability, which is the operating experience. The data of Table 2 characterise reliability of the clad valve components. These data were obtained on the basis of generalised experience of operation of main steam gates of the 300 MW units in service at heat power plants. It was found that the gate components with D_n 150 mm, clad using electrodes TsN-2 and TsN-12, had a mean life to failure (to formation of cracks on the seal surfaces) equal to 4.08–20.40 thous. h, which is much lower than the design life of the main power generation equipment. The best performance of the seal surfaces is provided by plasma cladding using alloys of the Ni–Cr–Si–B system. In this case the mean life to failure is 46.7 thous. h for plates and 210 thous. h for seats.

Appearance and macrosection of clad parts are shown in Figure 13.

Good bead formation provides reduction of labour consumption in machining due to decrease in allowances. Thickness of the deposited layer is usually 3.3–3.6 mm, and after machining it is $2^{+0.5}$ mm. In manual and automatic submerged-arc cladding the deposited layer has thickness of 10–12 mm.

Application of plasma-powder cladding for mass production of parts of power generation, Christmas tree and ship valves [1, 8, 12] allowed the process productivity to be raised from 2 to 3 times, quality of the deposited metal to be fundamentally improved, and performance of the treated parts to be substantially increased.

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COBALT-BASED ALLOYS FOR SURFACING AGAINST WEAR AND CORROSION

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Cobalt-based alloys for resisting wear and corrosion were first developed prior to 1907 under the trademark Stellite®. Additional alloys were later developed for needs in a variety of applications. Cobalt alloys can be applied on the surface of industrial components by hardfacing or thermal spraying. This paper is intended to review the traditional, as well as recently developed cobalt alloys in regards to their metallurgical characteristics and industrial applications. Various technologies for surfacing are discussed. Examples of using cobalt alloys in various industries for surfacing are illustrated in this work.

Keywords: *cobalt alloys, Stellite, wear, corrosion, PTA, HVOF, valves, oil drilling, galvanizing, refinery, woodcutting*

Nearly 100 years ago, in 1907 [1], cobalt alloys were developed for use in food processing, dental, cutlery, machine tools and engine components. These alloys take the advantage of cobalt having substantial solid solubility for other elements, such as chromium, molybdenum and tungsten. In addition, carbides precipitate readily in the matrix. Co–Cr–W alloys were found to be not only hard and tough but also bright and shinny. Therefore, the trademark «Stellite» (star-like) was invented (Stellite® is a registered trademark of Deloro Stellite Holdings, Inc.). Later, welding consumables of these alloys were made. They were found to be friendly to surface welding, which lead to the term «hardfacing». Works [2, 3] give detail descriptions of hardfacing alloys and methods of applying them. In more recent times, additional alloys have been developed to meet the requirements of many demanding industrial applications. Also, with the advent of modern surfacing technologies, cobalt alloys are used increasingly in many demanding applications in a variety of industries.

Table 1. Type 1 cobalt alloys

Alloy	*Nominal composition*	*Hardness, HRC*
Stellite® 6	29Cr–4.5W–1.2C	42
Stellite 12	29Cr–8.5W–1.5C	45
Stellite 1	29Cr–12W–2.4C	48
Stellite F	27Cr–22Ni–27Cr–1.8C	42
Stellite 190	27Cr–14W–3.2C	55
Stellite 694	28Cr–20W–5Ni–1V–0.9C	51
Stellite 706	29Cr–4.5Mo–1.2C	42
Stellite 712	29Cr–8.5Mo–1.5C	46
Stellite 703	32Cr–12Mo–2.4C	54
Stellite 720	33Cr–18Mo–2.5C–0.3B	60

Note. Stellite® is a registered trade mark of Deloro Stellite Holdings, Inc.

The properties of cobalt alloys depend on their microstructures, which can be categorized into three distinctive types, namely, type 1 – carbide, type 2 – intermetallic and type 3 – solid solution. The original cobalt alloys are type 1, type 2 alloys utilize the precipitation of intermetallic compounds for hardening and type 3 alloys are similar to type 1 but have minimum precipitation of carbides. Solution strengthening contributes to the strength of the alloys. In addition, they exhibit strong work hardening characteristics and very good corrosion resistance.

Type 1. Carbide alloys. As listed in Table 1, chromium and carbon are inevitably present in these alloys because they form the needed «chromium carbides» for wear resistance. They are typically in the form of M_7C_3 and $M_{23}C_6$ [4]. The M_7C_3 carbide, found mostly in hyper-eutectic alloys, has a higher melting point and precipitates early during solidification. The $M_{23}C_6$ carbide tends to precipitate later in the eutectic region of hypo-eutectic alloys after the formation of Co–Cr solid solution dendrites. Figure 1 illustrates these microstructures. The M_7C_3 carbide size is ≈ 100 μm. The $M_{23}C_6$ carbide particles are in the eutectic region.

Other elements, such as tungsten and molybdenum, are added to these alloys for improving other properties. Tungsten is known to partition in the carbides but not significantly due to the slow kinetics of tungsten. Molybdenum, on the other hand, is a smaller atom and has a strong affinity to carbon. The carbides formed by tungsten and molybdenum tend to be M_6C [4]. Under scanning electron microscope, it can be clearly seen that the Cr-rich and Mo-rich carbides are separate as illustrated in Figure 2.

Other than improving the mechanical properties of type 1 alloys, both tungsten and molybdenum also enhance corrosion resistance, especially in reducing or non-oxidizing conditions. Especially molybdenum has a strong effect to resist corrosion in reducing acids and complex environments. However, it weakens the corrosion resistance in purely oxidizing acids. In a

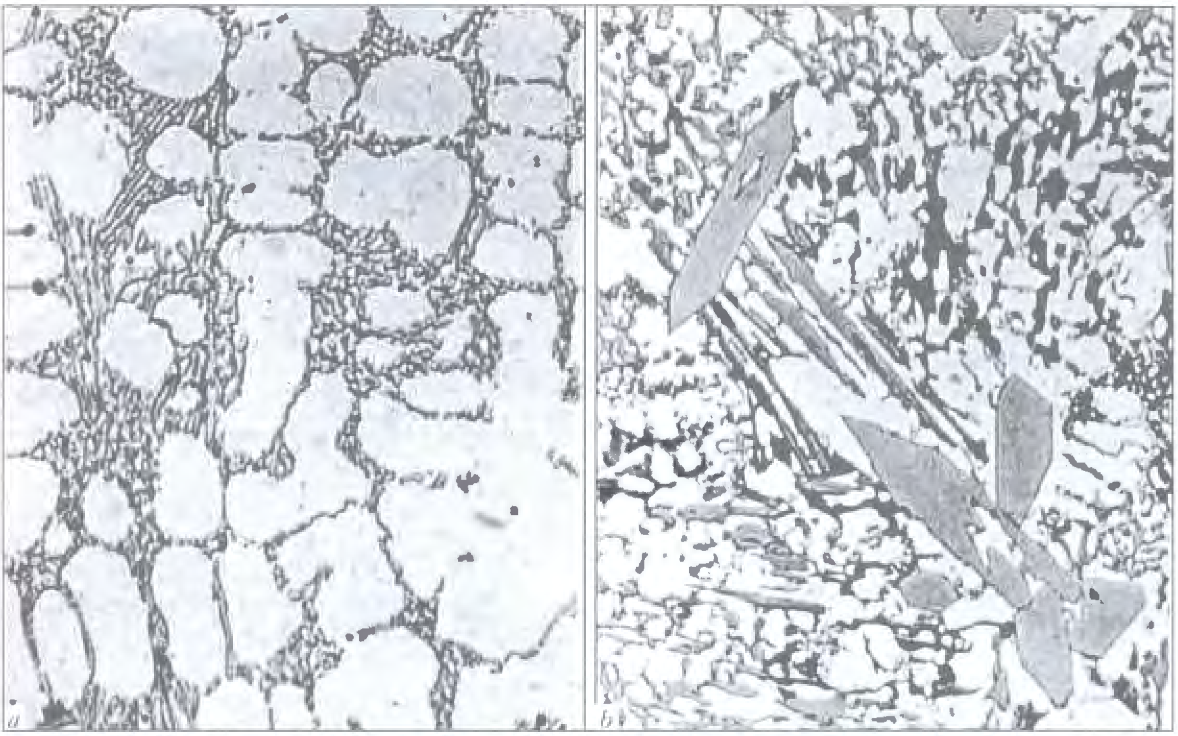

Figure 1. Hypo-eutectic structure with white Co–Cr solid solution dendrites (*a*) and hyper-eutectic structure with elongated large M_7C_3 carbide particles (*b*) of type 1 cobalt alloys

complex pitting environment, molybdenum has a strong beneficial effect as evidenced in the commonly adopted pitting resistance equivalent number formula for iron and nickel alloys. Table 2 compares the alloys containing either tungsten or molybdenum in different acids. It can be clearly seen that the Mo-containing alloys are more corrosion resistant than their W-containing counterparts in the reducing acids.

Type 2. Intermetallic alloys. In 1970s, a family of alloys was developed for resisting not only wear but also corrosion by chemicals. These alloys contain a high molybdenum content, as well as a high silicon content. These two elements contribute to the formation of an intermetallic phase called Laves phase, which has been identified to be within the stoichiometric limits of Co_3Mo_2Si and CoMoSi [5], having an A_2B crystal structure. Figure 3 shows a typical microstructure for a type 2 alloy. The Laves phase shows up as «flower-like» precipitates. It renders the type 2 alloys high hardness values and is thermally stable at high temperatures. As seen in Figure 4, the hardness maintains at a higher level for a type 2 alloy than a type 1 alloy as temperature in-

Table 2. Corrosion test results of type 1 alloys in acids

Alloy nominal composition	*Corrosion rate, mm/year*		
	5 % HCl at room temperature	*10 % HCl at room temperature*	*10 % H_2SO_4 at 66 °C*
29Cr–4.5W–1.2C	2.40	3.00	43.0
29Cr–4.5Mo–1.2C	1.00	0.33	9.7
32Cr–12W–2.4C	0.78	1.20	25.0
32Cr–12Mo–2.4C	0	0.23	0

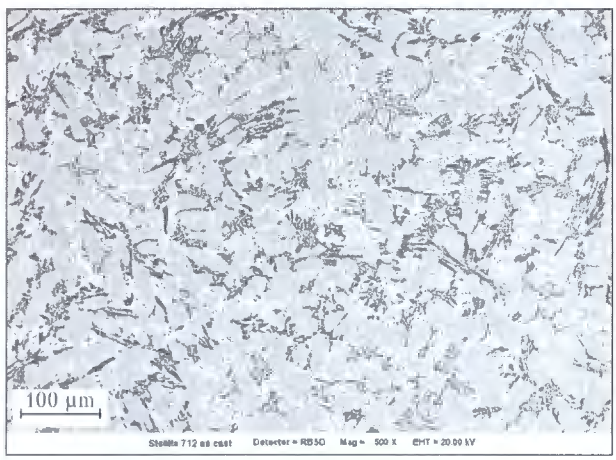

Figure 2. Scanning electron micrography of Stellite 712 (the dark and white areas contain, respectively, Cr-rich and Mo-rich carbide particles)

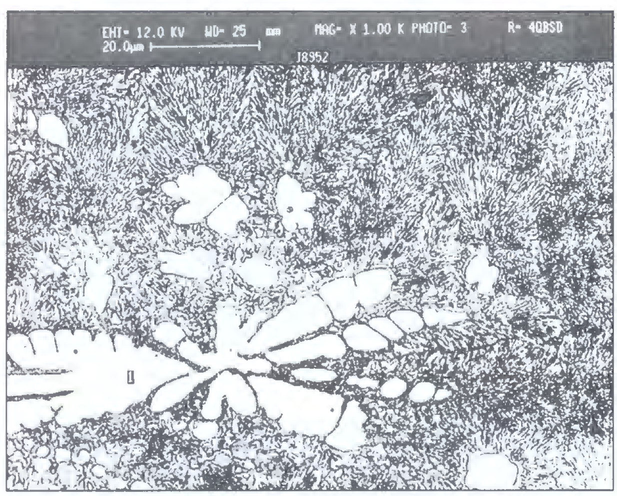

Figure 3. Microstructure of a type 2 alloy (Tribaloy T-900) showing Laves phase

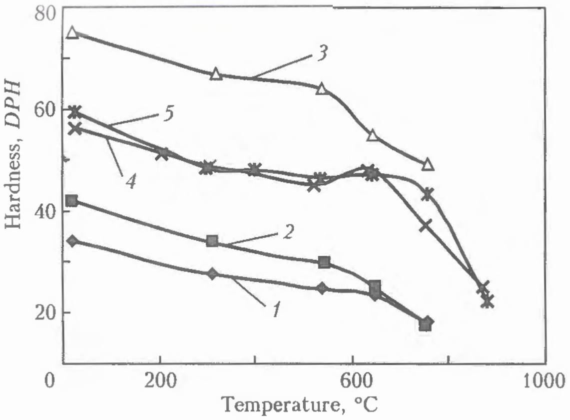

Figure 4. Hardness as a function of temperature for types 1 and 2 alloys: *1* — Stellite 21; *2* — Stellite 6; *3* — Tribaloy T-800; *4* — Tribaloy T-400; *5* — Tribaloy T-400C

creases. Many high temperature industrial applications utilize this characteristic for solving wear problems. Table 3 lists the commercially available type 2 alloys.

Laves phase is known to be brittle. Therefore, care must be taken in using type 2 alloys. For example, substantial preheating is necessary before weld overlaying. Recently, a proprietary alloy with a hypo-eutectic Laves microstructure has been developed for crack-free welding on a large surface. This alloy has been named Tribaloy T-401.

Type 3. Solid solution alloys. These alloys are basically the low-carbon version of type 1 alloys. Because of the minimum precipitation of carbide particles, the properties are more dependent on the matrix than type 1 alloys. Cobalt is known to go through a phase transformation from the γ-phase having a face-centered cubic (fcc) structure to the ε-phase having a hexagonal close pack (hcp) structure during solidification. However, the kinetics of this transformation is slow. This leads to predominantly fcc phase in cobalt alloys at room temperature, especially for weld overlays. Under stress, the fcc phase tends to transform to the hcp phase [6]. Such transformation absorbs the energy from the stresses and contributes to the wear resistance. Type 3 alloys are known to work harden significantly. For example, Stellite 21 can work harden from *HRC* 28 to 45. This characteristic is especially useful in cavitation erosion situations, where absorption of the shock wave energy from implosion of water bubbles is important.

Due to the low content of carbide precipitates, the corrosion resistance of type 3 alloys is generally good. Carbide formation reduces the amounts of chromium and molybdenum in the matrix, which are needed for corrosion resistance.

Table 3. Type 2 cobalt alloys

Alloy	*Nominal composition*	*Hardness, HRC*
Tribaloy® T-400	28Mo–9Cr–2.6Si–0.04C	55
Tribaloy T-400C	27Mo–14Cr–2.6Si–0.08C	55
Tribaloy T-800	28Mo–17Cr–3.4Si–0.04C	58
Tribaloy T-900	23Mo–17Cr–16Ni–2.7Si–0.04C	54

Note. Tribaloy® is a registered trademark of Deloro Stellite Holdings, Inc.

Table 4. Type 3 cobalt alloys

Alloy	*Nominal composition*	*Hardness, HRC*
Stellite 21	28Cr–5Mo–0.25C	28
Ultimet®	25Cr–9Ni–5Mo–2W–0.05C	27

Note. Ultimet® is a registered trademark of Haynes Int. Company.

Two commercially available type 3 alloys as listed in Table 4. Additional alloys are under development for specific applications.

Surfacing consumables and methods. There are a number of ways to apply cobalt alloys onto a metallic surface for protection against wear and corrosion. The consumables and methods are tabulated in Table 5. Many of the traditional welding methods can be used for applying cobalt alloys, as well as the more modern welding methods. Especially, plasma transferred arc (PTA) hardfacing with Stellite powder became a state of the art in manufacturing of industrial valves and knives, engine valves, plastic extrusion screws, and wear plates. Newest trends in PTA applications are manual and robotized hardfacing of glass moulds and small gas turbine blades with type 1 Stellite alloys.

It should be noted that the microstructures and, therefore, properties of the overlays are often dependent on the application methods [7]. For weld overlays, dilution with the substrate is an important factor for the properties. Different welding methods may result in significantly different distribution of dilution with the substrate metal in the weld overlays. For example, Figure 5 shows that a PTA weld overlay of Stellite 12 has iron dilution limited to only 0.06 mm from the bond line whereas a TIG weld overlay of a 3.2 mm

Table 5. Surfacing consumables and common application methods

Consumables	*Application method*
Wire (solid or tubular)	MIG, TIG, submerged-arc and open-arc
Powder	PTA, laser cladding, powder welding, HIP cladding, HVOF, plasma and flame spraying
Cast rod	TIG, oxy-acetylene, electrical resistance
Coated electrode	Manual MIG

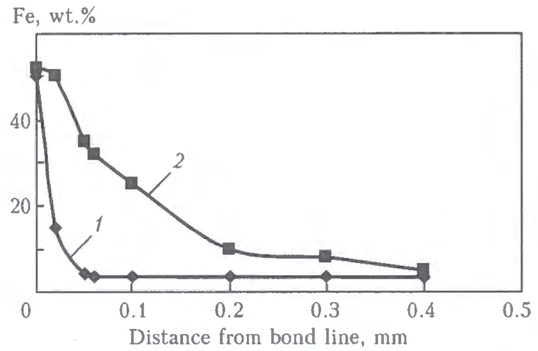

Figure 5. Iron content in weld overlays versus distance from bond line showing different distributions of dilution: *1* — PTA weld; *2* — TIG weld

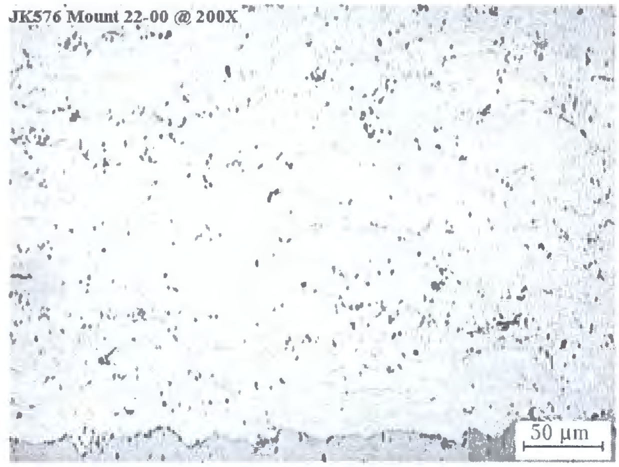

Figure 6. Microstructure of Stellite 6 coating sprayed with Jet Kote HVOF system (Jet Kote® is a registered trademark of Deloro Stellite Holdings, Inc.)

rod has evidence of iron dilution even at 0.4 mm from the bond line.

Adding carbon to the weld deposits by the carburizing flame of an oxy-acetylene torch can alter the original chemical composition. The high cooling rates in laser cladding could result in many interesting features in cobalt alloys [8].

When cobalt alloys are applied by high velocity oxy-fuel (HVOF) thermal spraying, the microstructure is entirely different from weld overlays as seen in Figure 6. The coatings consist of rapidly quenched splats, as well as oxide inclusions, which can obviously influence the coating properties.

Industrial applications. The traditional type 1 alloys are widely used in non-lubricating applications. For example, in diesel engine valves, hardfacing on the valve trim with Stellite 6, F or 12 alloy is a common practice as illustrated in Figure 7.

Figure 7. PTA hardfaced engine valves with a type 1 Stellite alloy

In chemical processing, hardfacing on the valve trims or seats with a type 1 alloy is widely used to prolong the service life as shown in Figure 8. In oil drilling, the bearing area of a tri-cone drill bit is typically hardfaced with Stellite 190, a hyper-eutectic type 1 alloy. Other drilling components are often hardfaced with a type 1 alloy as in the case of a piston shown in Figure 9. Stellite alloys have also found many applications in the oil refineries since early years [9].

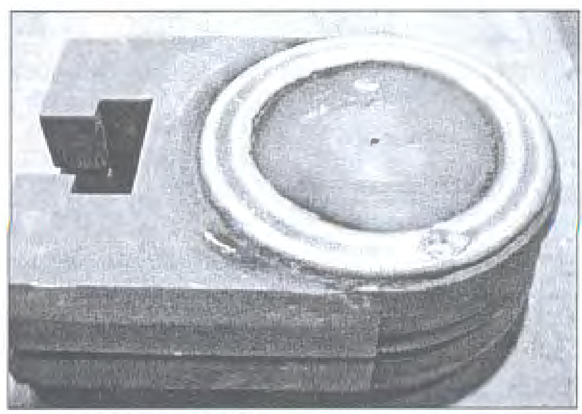

Figure 8. PTA hardfacing with a Stellite alloy on a valve seat

In woodcutting, different type 1 alloys can be used to tip the saw blades for maintaining the sharpness of the teeth because of their resistance to abrasion and corrosion by the wood juices. Figure 10 shows tipping a saw blade with Stellite 12. In recent years, new Stellite alloys have been developed for further improving the performance [10].

In gas turbines for either aerospace or power generation, Stellite 694 wire is used routinely to hardface the Z-notches at the top end of a turbine blade for protection against fretting. As the design temperature for land-based turbines is increasing, alloy selection is gradually leaning toward the more stable type 2 alloys.

Type 2 alloys are often used when the operating temperature is too high for the type 1 alloys. Examples are diesel engine valves and exhaust gas circulation (EGR) valves in automobiles. Certain petrochemical processes also require the properties of type 2 alloys.

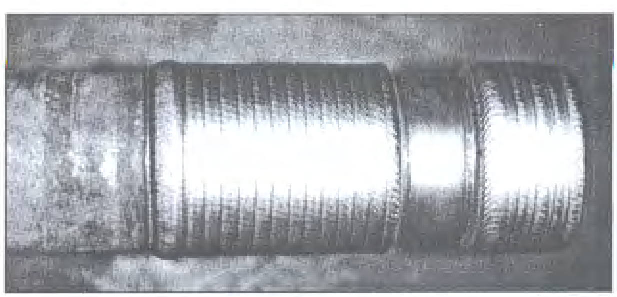

Figure 9. Oil drilling piston hardfaced with Stellite 712 by PTA

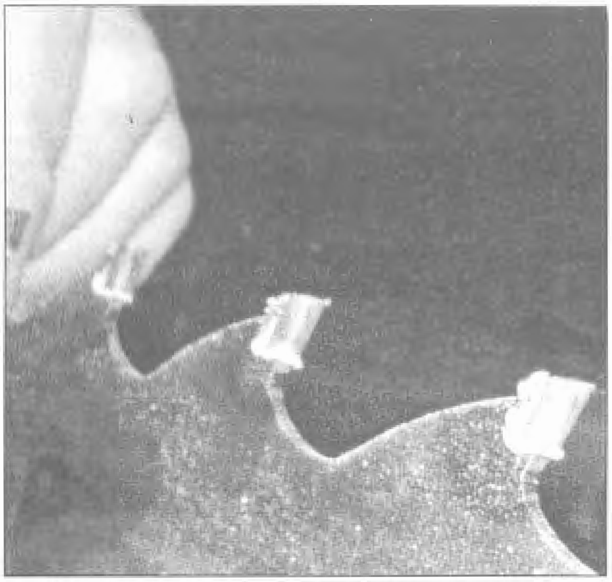

Figure 10. Saw tipping with a type 1 Stellite alloy

Type 2 alloys are also found resistant to galvanizing bath (continuous hot dipping of steel sheets in molten zinc with a small amount of aluminum). This property makes it useful in hardfacing bearings for both the sink rolls and stabilizing rolls in continuous galvanizing. Hot metalworking is another industry where type 2 alloys can be used.

Type 3 alloys are used for hardfacing on hydroturbine blades to minimize cavitation erosion. They can also be used on valves where the relative movement of the wear couple is mostly pressing not sliding. The excellent saltwater corrosion resistance makes these alloys suitable for wear applications in saltwater environments. One of the type 3 alloys, Stellite 21, has been used for hardfacing hot forging dies for many years due to its suitable properties under the operational conditions.

CONCLUSIONS

1. Cobalt alloys for surfacing can be categorized into three types based on their microstructures. Each type has its own characteristics as summarized below:

• Type 1 alloys have a microstructure of precipitated Cr-rich carbide particles, which contribute to the wear resistance. Tungsten and molybdenum tend to form a different kind of carbide.

• Type 2 alloys contain an intermetallic phase, which is stable at high temperatures.

• Type 3 alloys show strong work hardening characteristics and excellent corrosion resistance.

2. All three types of alloys have proven performance in many industrial applications as shown in the examples given in this paper.

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PLASMA-POWDER SURFACING OF COMPOSITE ALLOYS BASED ON CAST TUNGSTEN CARBIDES

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Wear resistance of deposited metal, depending on the quantity and shape of the reinforcing tungsten carbide particles, methods for their addition to the weld pool and type of matrix has been investigated. It is shown that the best combination of service and technological properties of the deposited metal is achieved at its carbide content of about 50 vol.%.

Keywords: *plasma surfacing, filler powder, spherical tungsten carbide, matrix, wear resistance, carbide distribution, hardness, surfacing equipment*

For surfacing parts operating under intensive abrasive wear, the most effective are composite alloys based on cast tungsten carbides (further on referred to as tungsten carbide) [1, 2]. Despite the high cost, they are quite often irreplaceable, particularly in mining industry. Composite alloys are successfully used for surfacing such parts, as drill pipe locks, drill bit roller cones, dipper teeth, etc. [3–5]. The service life of parts surfaced with these alloys, is several times higher than that of parts surfaced with hypereutectic alloys of the type of high-chromium cast irons.

In this connection searching for effective processes of surfacing composite alloys, providing the best combination of the deposited metal performance, is highly urgent. Plasma-powder surfacing (PTA-surfacing) should be noted among the known technological process, applied for surfacing composite alloys [6].

This study is devoted to investigation of wear resistance of deposited metal, depending on the amount and shape of reinforcing particles of tungsten carbide, method of their addition to the weld pool, as well as type of binder-alloy (matrix).

Used as the binder-alloy in the experiments, were powders of Ni- and Fe-base alloy (Table 1). Powder fraction was 56–200 μm. Selection of these alloys was not accidental. Ni-based self-fluxing alloys are widely used for surfacing composite materials. They are characterized by a comparatively low melting temperature (1000–1100 °C), good wettability of tungsten carbide grains and sufficient wear resistance. Alloys differing essentially by hardness and ductility were selected for our experiments.

Fe-based alloy of the above composition was suggested for these purposes for the first time. It belongs to the class of high-vanadium cast irons and combines a high wear resistance with ductility, required to ensure the working layer resistance to impact loads. In addition, it is much less expensive than Ni-base alloys.

Powders of crushed and spherical tungsten carbide co-produced by «Toreztverdosplav» Plant and E.O. Paton Electric Welding Institute were used as reinforcing materials. Sizes of these powder particles were 100–300 μm.

Surfacing was performed on samples of 60 × 100 × 20 mm size of steel 20 with RR-6-02 plasmatron in the equipment of Plasma-master, Ltd. (Kiev). Section of deposited beads was equal to 28 × 5 mm, surfacing efficiency was 5 kg/h.

Matrix and reinforcing powders were fed using two processes, namely as a mixture and separately. In the first case, in order to avoid separation, the powders were mixed by coalescence of two flows, fed from two separate feeders directly before entering the plasmatron; in the second case the matrix alloy powder was fed into the arc through a focusing nozzle of the plasmatron in the form of a distributed flow, and tungsten carbide — directly to the weld pool by a special channel (Figure 1). Such a method of feeding the particles of tungsten carbide into the weld pool minimizes the arc impact on them, thus preserving them from dissolution. This is particularly important, when Fe-based alloys are used as the matrix [2, 3].

Table 2 presents all the tested variants of deposits. The separate feed method was used predominantly for

Table 1. Composition of alloys used as the binder-metal

Alloy type (matrix)	*Element content, wt.%*									*Hardness*
	C	*Si*	*B*	*Ni*	*Fe*	*V*	*Mo*	*Cr*	*Cu*	
Ni–Cr–Si–B	0.50	2.6	2.2	Base	2.1	–	–	13.5	–	*HRC* 40
Ni–Cu–Si–B	0.15	1.1	1.0	Same	1.8	–	–	–	42.5	*HB* 200
Fe–Cr–V–Mo–C	2.20	0.6	–	2.7	Base	7.8	2.5	18.2	–	*HRC* 44

Table 2. Deposit variants and evaluation of wear resistance

No. of deposit variant	*Alloy type (matrix)*	*Volume fraction of tungsten carbide, %*	*Method of powder feed*		*Shape of tungsten carbide particles*		*Presence of cracks*	*Loss of sample mass at testing, g*	
			Mixture	*Separate*	*Spherical*	*Fragments*		*FR*	*G-65*
1	Ni–Cu–Si–B	40	X		X		+	0.092	–
2	Same	40	X			X	+	0.081	–
3	»	40		X	X		+	0.098	–
4	»	50	X		X		–	0.035	0.038
5	»	50	X			X	–	0.052	0.059
6	»	50		X	X		+	0.034	–
7	»	50		X		X	+	0.046	–
8	»	60	X		X		+	0.079	–
9	»	60	X			X	+	0.092	–
10	»	55		X	X		–	0.028	0.036
11	»	55		X	X		–	–	0.037
12	Fe–Cr–V–Mo–C	50	X		X		+	0.055	–
13	Same	55		X	X		–	0.026	0.029

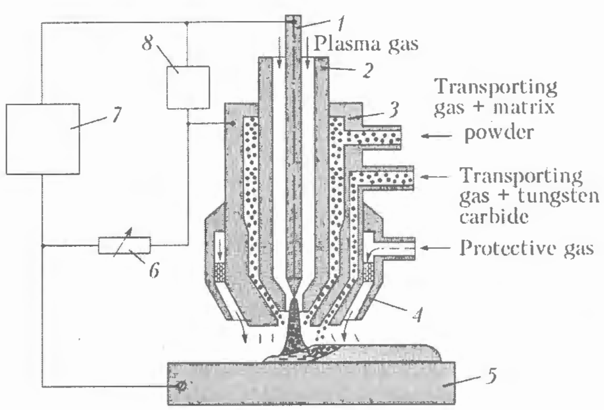

Figure 1. Schematic of plasma-powder surfacing of composite alloys with a separate feed of matrix and reinforcing materials: *1* — electrode; *2, 3* — plasma and focusing nozzle, respectively; *4* — protective nozzle; *5* — part; *6* — ballast rheostat; *7* — power source; *8* — oscillator

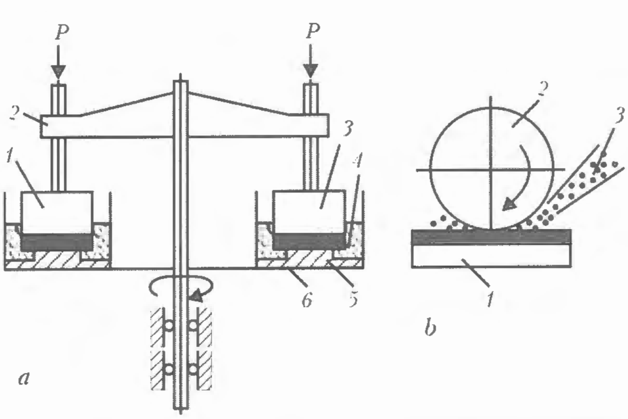

Figure 2. Schematic of testing samples for wear resistance by different procedures: *a* — FR (*1* — tested sample; *2* — rotating cross-piece; *3* — standard; *4* — water with abrasive; *5* — copper ring; *6* — vessel); *b* — G-65 (*1* — tested sample; *2* — rotating rubber disc; *3* — dried quartz sand)

spherical tungsten carbide as the most promising material for this process.

Surfaced plates were used to cut out samples of 16 × 6 × 16 mm size for wear resistance testing by «fixed ring» (FR) procedure [7] and of 76 × 25 × × 12 mm size for wear resistance testing by a standard ASTM G-65 procedure (dry sand–rubber disc). The same samples were also used for metallographic investigations. Upper part of the deposited metal layer was polished to the level at which the tungsten carbide grains were located relatively uniformly across the bead section.

The first type of samples were tested at PWI, the second in the Research Laboratory of Alberta Research Council Inc. (Edmonton, Canada). In the second case those variants of deposits were tested, where the best carbide distribution was found. This procedure envisages two testing stages, the first being running-in (6000 rpm of the rubber disc), and the second being the test (6000 rpm on the run-in surface). Schematics of sample testing are shown in Figure 2, testing results — in Figure 3.

Discussion of results. *Surfacing with a powder mixture*. The best results were achieved at the content of tungsten carbide in the mixture of about 50 vol.%

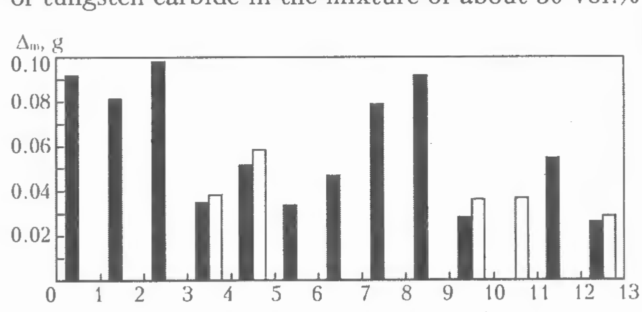

Figure 3. Diagram of wear resistance of different deposit variants in testing by FR (■) and G-65 (□) procedures: Δ_m — loss of sample mass

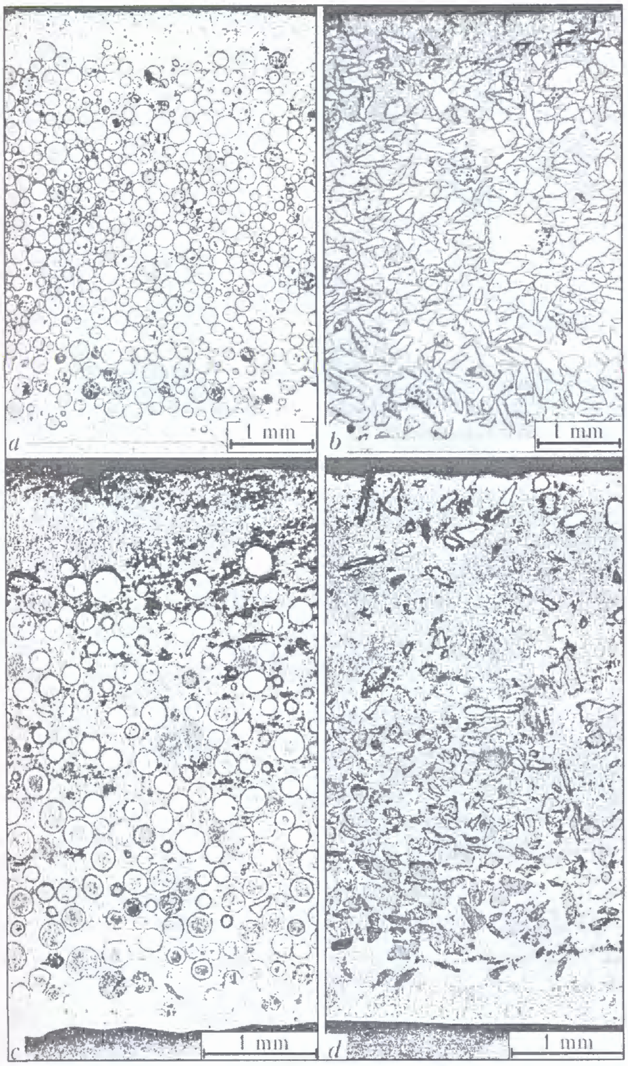

Figure 4. Distribution of spherical tungsten carbide (*a*, *c*) and crushed (*b*, *d*) tungsten carbide in the deposited metal at its content of 50 (*a*, *b*) and 60 (*c*, *d*) vol.% in the mixtures (matrix is Ni–Cr–Si–B alloy)

(variants of deposits 4, 5). This is true both for the spherical and crushed tungsten carbide, irrespective of the matrix type. In this case, a good formation of the beads and uniform distribution of tungsten carbide particles over the section are observed (Figure 4, *a*, *b*), thus providing maximum wear resistance of the deposited metal (see Figure 3). If the volume fraction of tungsten carbide in the mixture is greater than 50 vol.% (variants 8, 9), then for a good formation of the bead it is necessary to significantly (by 40–50 %) increase the surfacing current, and, hence, also the heat input into the part. This is due to insufficient amount of the low-melting component in the powder mixture (alloy-binder), which leads to a marked dissolution of particles, matrix embrittlement, and, consequently, to lowering of the wear resistance of the deposited metal (see Figure 3). The molten metal retains not more than 30 vol.% of spherical (Figure 4, *c*) and not more than 25 vol.% of crushed tungsten carbide (Figure 4, *d*).

In the case of tungsten carbide content in the mixture below 50 vol.% (variants 1, 2), all of it is located

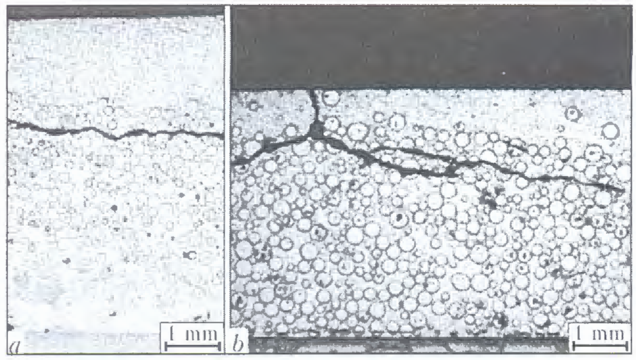

Figure 5. Crack location in the metal layer at incomplete filling with tungsten carbide deposited using a mixture (*a*), and separate powder feed (*b*)

in the lower part of the deposited metal layer, leaving its upper part unfilled. It is natural that the wear resistance of the latter, despite the additional alloying with carbon and tungsten remains to be low (see Figure 3). Results of studying the distribution of matrix microhardness by the height of the deposited metal and of X-ray microprobe analysis showed that carbide dissolution proceeds in all the cases, although to varying degrees, being more noticeable in the upper part of the deposited metal layer, namely in the zone of the plasma arc impact. Even in variants 4 and 5 with 50 vol.% of tungsten carbide, microhardness of the matrix solid solution due to its alloying with carbon and tungsten rises from HV_{01} 366 Pa (at the fusion line) up to HV_{01} 727 Pa (in its upper part).

Microhardness of well-preserved particles of spherical and crushed tungsten carbide is about the same in the alloy and by the data generated in Otto-von-Guericke-Universitat Magdeburg (Germany), it is HV_{01} 3900–3991 Pa, while the microhardness of half-dissolved particles is in the range HV_{01} 2500–2650 Pa.

In all the studied variants of the deposits, microcracks were present in the deposited metal. Their smallest number was observed in variants 4 and 5, and the greatest number in variants 8, 9 and 12. In variants 1 and 2, where the bead upper part is not filled with the tungsten carbide, the cracks were located along the layer interface (Figure 5, *a*). When spherical tungsten carbide is used, also spallation of the deposited metal is found because of high internal stresses.

Surfacing with separate powder feed. Dissolution of tungsten carbide at separate feed of the matrix and reinforcing materials is minimum, even in the case of

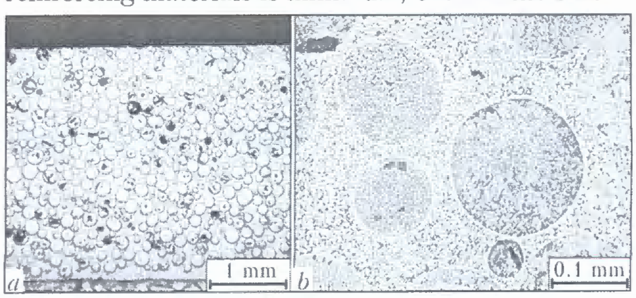

Figure 6. Distribution of tungsten carbide grains in the iron-base matrix at separate powder feed: *a* — ×20; *b* — ×100

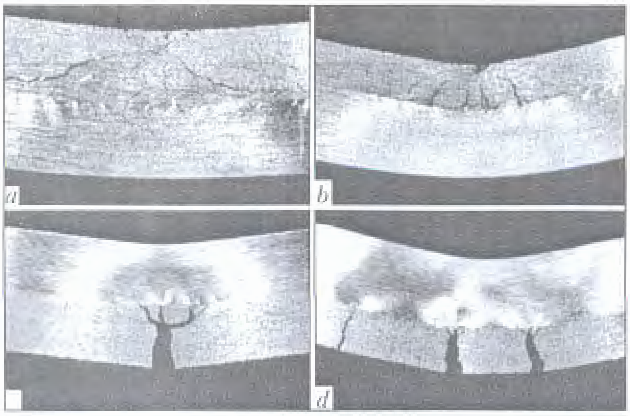

Figure 7. Bend testing of samples surfaced with a separate feed of spherical tungsten carbide: *a, c* — Ni–Cr–Si–B matrix; *b, d* — Fe–Cr–V–Mo–C matrix

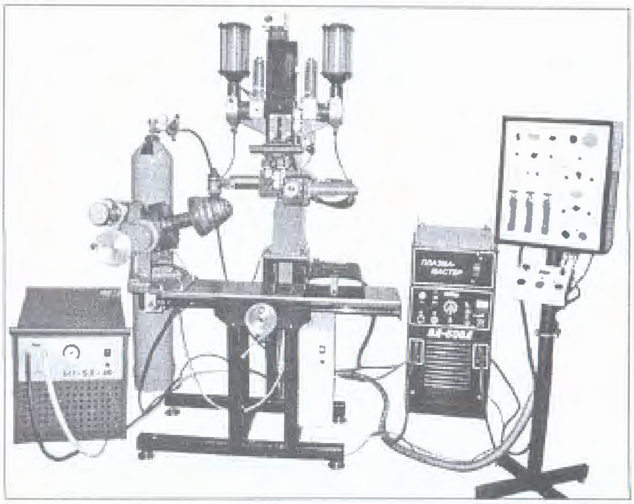

Figure 8. Unit for plasma-powder surfacing with a separate feed of PM-300S powders

an Fe-based matrix (variant 13, Figure 6, *a*). This is confirmed by the data of X-ray microprobe analysis and measurement of matrix microhardness. Microhardness of solid solution of Ni–Cr–Si–B matrix was HV_{01} 360–420 and that of Fe–Cr–V–Mo–C system was HV_{01} 540–640. The beads were well formed and cracks were absent in the deposited metal layer in most of the cases. Similar to surfacing with a mixture, they initiated in those cases, when a partial filling of the bead volume with the carbides was observed (variant 3, Figure 6, *b*).

Value of wear resistance of the deposited metal with a matrix of Ni–Cr–Si–B system at its uniform filling with tungsten carbide (variant 10) is a little higher than that which was obtained in surfacing with a mixture (see Figure 3). Wear resistance is also higher in the case of using a softer nickel matrix of Ni–Cu–Si–B system (variant 11). It is obvious that matrix hardness has only a slight effect on the total wear resistance of the deposited metal layer, and this means that tungsten carbides have a prevailing role under such conditions of wear. It should be also noted that absence of cracks in the deposited metal layer, when using powder of Ni–Cu–Si–B system makes it a quite promising material for these purposes.

When an Fe-base alloy is used as the matrix, the deposited metal has an even higher wear resistance (variant 13, Figure 3). It should be noted, however, that in surfacing with a separate powder feed, losses of tungsten carbide become greater as part of it does not get to the weld pool. In order to achieve a uniform filling of the deposited metal layer with tungsten carbide, it is necessary to compensate the tungsten carbide losses by increasing its feed relative to matrix powder by 5–8 %.

At other conditions being equal, wear resistance of composite layers of metal with spherical tungsten carbide is mostly much higher than that of layers with the crushed tungsten carbide, contrarily to conclusions of [8]. Spherical tungsten carbide is dissolved in the weld pool to a smaller degree, and it is very convenient for plasma-powder surfacing, as it has a higher yield point, and does not cause equipment wear.

Bend testing of deposited samples with load application both from the side of the deposited (Figure 7, *a, b*), and the side of base metal (Figure 7, *c, d*) showed that the layer of the deposited metal has a good adhesion to the base metal and does not delaminate. It is valid both for the Ni- and Fe-base matrix.

Figure 8 gives the equipment for plasma-powder surfacing by composite alloys, and Figure 9 shows

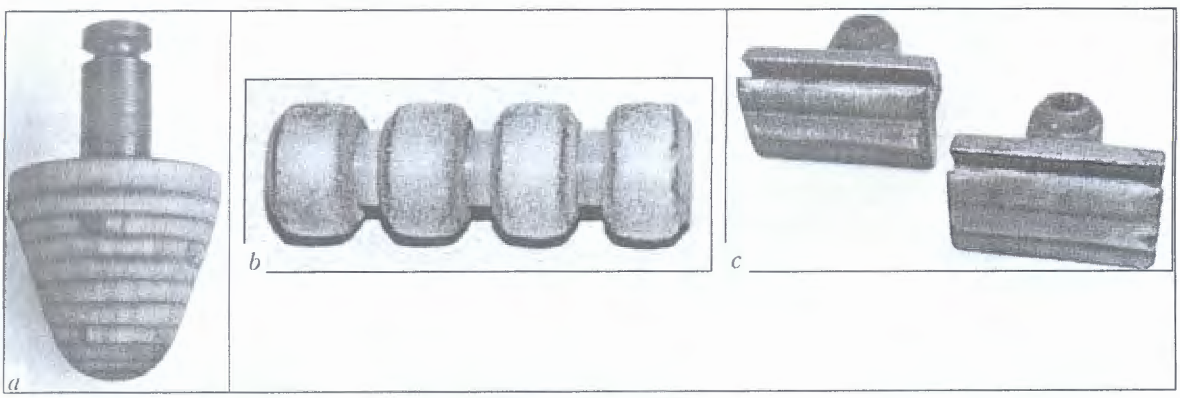

Figure 9. Examples of parts surfaced with composite materials with separate wire feed: *a* — crusher tooth; *b* — calibration rolls; *c* — crusher baffles

the parts surfaced with the above materials with a şeparate powder feed.

CONCLUSIONS

1. The best combination of service and technological properties of the metal both in surfacing with a mixture, and with separate feed of the powders is achieved at about 50 vol.% content of tungsten carbide in it.

2. In the case of a separate feed of powders, dissolution of spherical and crushed tungsten carbides is lower than in surfacing with a mixture.

3. In surfacing with a mixture of powders a matrix which is more ductile and more neutral to dissolution of matrix carbides, for instance, an alloy of Ni–Cu–Si–B alloying system, should be applied to avoid cracks and delamination of the deposited metal.

4. In the case of a separate feed of the matrix and reinforcing powders, iron-base alloys can be successfully used as the matrix.

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IMPROVEMENT OF TECHNOLOGY FOR HARDFACING OF METALLURGICAL EQUIPMENT COMPONENTS

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The ways of improving the quality of deposited metal through forming the working surface by layers of variable chemical composition and complex shape are considered. The latter increase the technological capabilities of the process, widen the sphere of application of deposited components and impart them a new combination of service properties.

Keywords: hardfacing, working layer, chemical composition, weld pool, equipment, technology

At different methods of surfacing the increase in service characteristics of the deposited working layer is attained, as a rule, by a proper selection and optimizing the chemical composition of electrode or filler materials and use of post (if necessary) heat treatment.

At the Priazovsky State Technical University and other organizations the works are carried our over the many years for the development of the new ways of improving the economic and service characteristics of hardfacing. The most challenging ways are as follows:

- hardfacing of layers with a preset distribution of properties depending on intensity and types of a part wear;
- control of sizes and path of deposition of beads on the part working surface.

As the most parts are worn out non-uniformly, their hardfacing with a layer of a similar chemical composition does not give the same results as the hardfacing of a layer with a distribution of chemical composition and properties depending on the intensity and types of wear [1–6]. In service of parts, strengthened by this way, a uniform or regular wear of the working surface is provided, that increases significantly their service life and characteristics, thus improving the quality of products being treated by them [7–9].

Control of deposited metal by alloying is a rather complicated problem, whose proper solution is possible only in obtaining of complete information about the current parameters of the weld pool: its dimensions and shape, chemical composition, share of parent metal participation in the deposited metal and others. As it is impossible to realize the direct measurements of these parameters, the problem of realization of technology of hardfacing of a working layer of a variable chemical composition can be solved by the development of different kinds of models [3, 10–13] describing the processes of alloying the weld pool and its formation, and also by the creation of the automated equipment for control of these processes using the above-mentioned modeling.

The most sophisticated procedures of calculation and algorithms of control of the process of formation of weld of a variable chemical composition are described in works [14–16], where such factors are taken into account as the effect of weld pool volume on the nature of changing the weld chemical composition, share of participation of parent metal and prior deposited layer in alloying, limitation in capabilities of alloying of the deposited metal at high gradients of changes in properties and others. They were laid down into the basis of system of the automatic control of the process of hardfacing of a working layer of a variable chemical composition [17, 18].

From the other side, the uncontrollable macro- and microchemical inhomogeneity of the deposited metal can also lead to the non-uniform wear of parts. This is illustrated by the formation of circular fire cracks in deposited mill rolls, rollers of machines for continuous casting of billets and other parts operating under the conditions of variable temperatures and pressures. Many researchers [19–21] associate the initiation of these cracks with specifics of the technology and technique of arc hardfacing of parts by a helical line with overlapping of neighboring beads.

Circular fire cracks on the deposited beads deteriorate the quality of the rolled metal. Moreover, deposited parts can fracture in sites of these defects.

To eliminate the above-mentioned drawbacks and to improve the service characteristics of the deposited layer, it is suggested to use the technological processes of arc hardfacing with a complex trajectory of the electrode displacement with respect to the surface being deposited. Moreover, during hardfacing the electrode is moved along a complex preset trajectory with a formation of zigzag-like, wavy beads or those, mutually-crossed by the definite law [22–25]. Mutual crossing of deposited beads, their orientation under an optimum angle to the action of maximum loads allow producing the working surface with high characteristics of wear and fire resistance, thus decreasing greatly the harmful effect of macro- and microheterogeneity of the deposited layer [22, 26].

Sometimes, for example, in manufacture of rolls of semifinishing mills it is necessary to make craters of definite shape and arrangement on their working

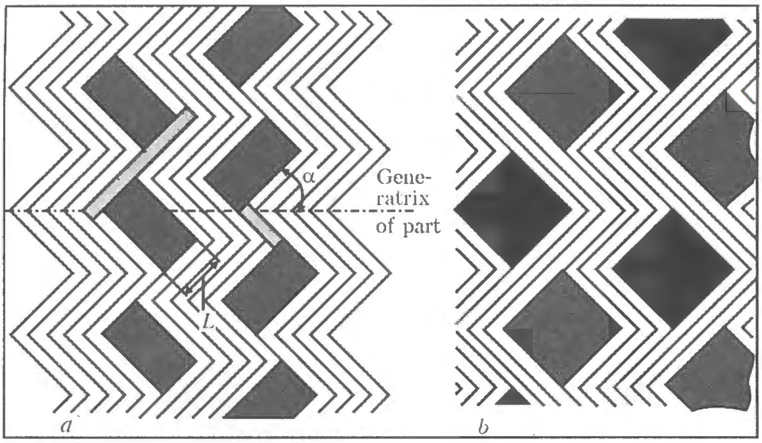

Figure 1. Schemes (*a, b*) of craters formation in hardfacing of working surface using the zigzag welds

surface to improve the adhesion and to increase the effectiveness of removal of scale from the rolled metal [27]. This operation is characterized by a high labour intensity, as the craters should be made on the surface having a high hardness. If to use the materials with a low hardness to facilitate the mechanical treatment, then it will influence negatively the service characteristics of the product.

Technological process of arc hardfacing of the working layer using beads of a zigzag shape (Figure 1) allows direct formation of a required relief on the working surface directly during hardfacing without additional mechanical treatment.

If in hardfacing of bodies of rotation along a helical line using zigzag beads with overlapping of neighboring (adjacent) beads at the moment of fulfillment of next straight-line area to increase its length by value L (areas with a changed length are conditionally distinguished in Figure 1, *a* by a grey color) and to deposit the zigzag weld with previous parameters of oscillation (amplitude and period) of the electrode end, i.e. to shift the oscillation axis along generatrix of a cylindrical part by $0.5L \cos \alpha$, then a series of tetragonal craters, located under similar angle to the part generatrix, are formed on the surface being deposited.

By decreasing the length of a straight-line area after several full rotations of the part relative to initial sizes for value L, we shall provide the formation of craters inclined under angle α with respect to the part generatrix but into opposite side relative to the first row of craters.

Number of full rotations of the part, after which it is necessary to shift the axis of electrode oscillation, defines the area of deposited regions between the rows of craters, i.e. the frequency of making craters in the longitudinal direction relative to the generatrix. Length L of change in extension of straight-line area of the zigzag weld defines the width of crater.

The similar effect can be attained by control of amplitude and period of oscillations in hardfacing of bodies of rotation along the helical trajectory using a zigzag weld, i.e. by changing them periodically and preserving the forms of oscillations and their phases (Figure 1, *b*).

Hardfacing of a working layer using welds of a complex shape is associated with the need in control of electrode displacement along the definite trajectory at a required accuracy. To realize the above-mentioned technology, the application of a specialized hardfacing equipment, optimum methods of calculation of configuration of welds and selection of parameters of their hardfacing, and also systems of automatic control of these parameters in accordance with a preset program are required.

In the process of designing the technology of deposition of a deposited layer with a complex configuration of arrangement of deposited beads, it is necessary to take into account and to optimize the large number of parameters. For example, in case of changing the trajectory of the electrode displacement in hardfacing with strip electrode, it is necessary to take into consideration, except other parameters, an angle of strip rotation relative to vector of hardfacing speed and to control it, as the specific heat input and depth of penetration, and also the quality of bead formation and, as a consequence, the serviceability of the deposited metal (resistance against selective and total wear, initiation and propagation of fire cracks and others) depend on it.

The more serious problems arise in hardfacing the cylindrical parts with shifting the bends of zigzag welds relative to a generatrix of cylinder (Figure 2), parts with a variable radius, for example, barrel- or cone-shaped parts. In this case it is necessary to change constantly the parameters of electrode oscillation and

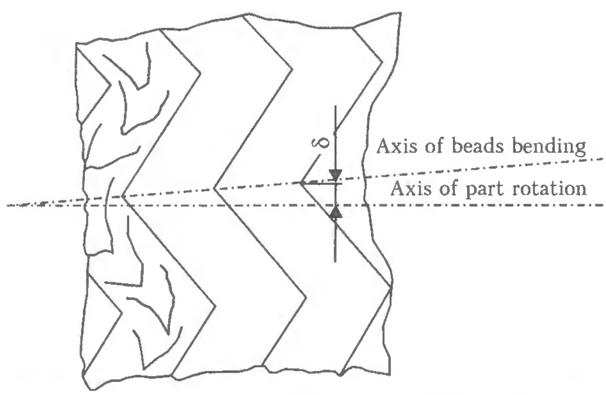

Figure 2. Scheme of hardfacing with shifting of zigzags apexes

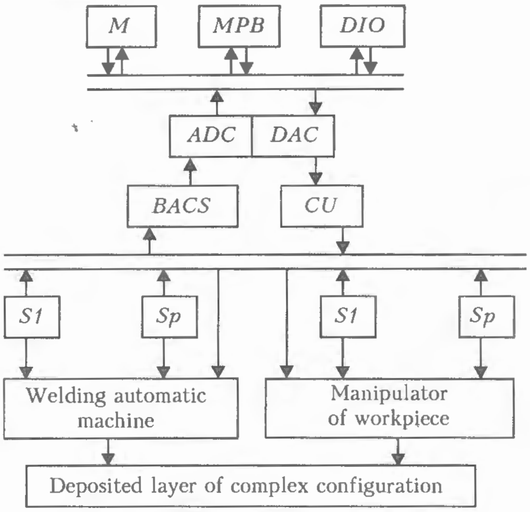

Figure 3. Functional scheme of system of automated control of hardfacing: *M* — monitor; *MPB* — microprocessor block; *DIO* — devices of input-output; *ADC* — analog-digital converter; *DAC* — digital-analog converter; *BACS* — block of amplification and conversion of input signals; *CU* — control unit; *S1*, *Sp* — sensors of position and condition parameters, respectively

to maintain simultaneously other parameters of hardfacing conditions at an optimum level.

The general description of the modified hardfacing equipment, allowing realization of electrode displacement along the required trajectory and having transducers giving information about coordinates relative to surface being deposited, is given in some works [16, 25, 27]. The main drawback of such equipment is its narrow specialization (impossibility to perform welds of different configuration as the change in its instrument part is required in this case).

The wide introdution of accessible, highly-reliable and universal microprocessor units into the system of monitoring and control allows solution of the above-mentioned problems using the proper software without changing and complication of the instrument part of the equipment.

The programmable control of the trajectory of electrode displacement gives opportunity to use successfully these systems for hardfacing products of a wide range of sizes and variants of configuration of the working layer, in particular, in cases when the quick resetting of the technological process is required.

The scheme of system of the automatic control of the trajectory of electrode displacement for realization of possible variants of hardfacing using welds of a complicated configuration is given in Figure 3. The system is equipped with a source of reference information and has feasibility of step-by-step control of the process of hardfacing of separate beads, that is especially important for the multiplayer hardfacing. The presence of electrode position feedback provides the maximum coincidence of calculation scheme with a real configuration of the deposited surface (Figure 4).

The use of the modified equipment at some metallurgical and machine-building enterprises in realization of technology of hardening by arc hardfacing of working surfaces of parts and tools of the equipment

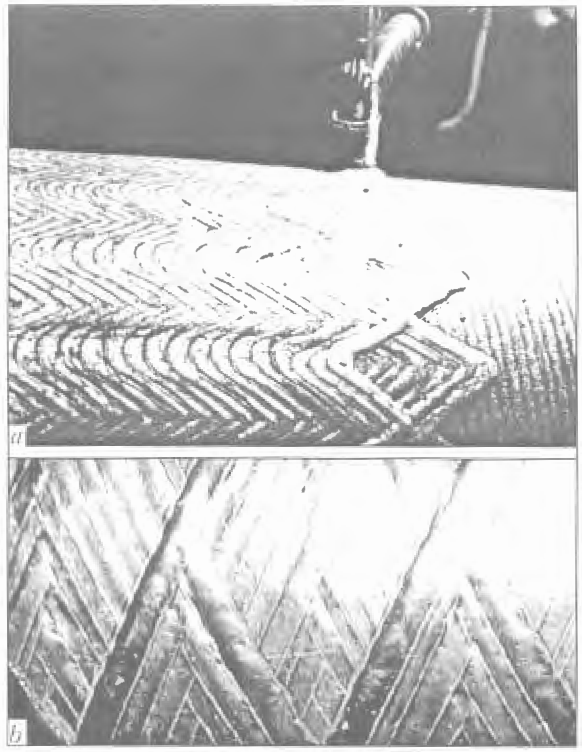

Figure 4. Appearance of working surfaces (*a*, *b*) deposited by beads of complex configuration

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ELECTROSLAG PROCESSING OF METAL WASTES AND USING OF RESULTING SEMI-FINISHED PRODUCTS FOR CLADDING APPLICATIONS

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Technologies developed by the E.O. Paton Electric Welding Institute for processing of different-production metal wastes by the electroslag process are considered. This approach holds promise in terms of involving secondary resources into production, as shown by examples of remelting and cladding using tool steel chips, slime wastes, ash residues and copper chops, and proved by positive results obtained from utilisation of the remelted metal in the form of finished or semi-finished products.

Keywords: *electroslag technologies, fine-fraction wastes, welding, remelting, cladding, spraying*

Wastes containing different metals, alloys and oxides of metals, such as chips of high-alloy steels and alloys, slimes left from grinding of parts of high-speed steels and special alloys, ash residues formed at heat power plants, etc., constitute a considerable part of the total volume of industrial wastes.

Processing of chips of tool steels, high-alloy die and high-speed steels in particular, addresses an important problem — recycling of a considerable amount of scarce and expensive alloying elements, such as tungsten, molybdenum, cobalt, etc. As reported in [1], chips resulting from machining of tool steel parts run 20–30 % of mass of an initial billet.

The problem of processing slime wastes is most difficult. These wastes include slimes formed in machining of special alloys based on nickel, cobalt, niobium and other elements [2, 3]. Such alloys are machined only by grinding, which results in 10–20 % of metal going to wastes. Abrasive dressing of semi-finished products and bars of high-speed steels results in about the same amount of losses [3, 4].

Industrial wastes include also slimes that fully consist of oxides of alloying elements. An example of such wastes is ash formed at heat power plants after combustion of high-sulphur residual oil [5]. This ash has a relatively high content of vanadium oxide (13–14 wt.%). Slimes can hardly be used in their initial state, and extraction of metal constituents from them involves much difficulties.

Different methods were suggested for recycling of the above wastes. For example, chips of alloyed steels and alloys can be remelted in electric arc furnaces to produce certified charge ingots, which are used as raw materials for high-quality metallurgy [6, 7]. However, substantial melting losses of alloying elements [8], depending in many respects upon the basicity and degree of deoxidation of slag [9], result from implementation of this technology.

It was suggested that processing of slimes could be performed by electrothermic methods [2, 10, 11], electric arc [12–14] and plasma [15] remelting, etc. Reported are the cases of using wastes of high-speed and other steels and alloys in coverings of welding electrodes [16–19], or as additions to powders for plasma cladding [20] and charge of surfacing flux-cored wire [21]. However, these methods of recycling of industrial wastes failed to find wide commercial application.

Ash residues of heat power plants can hardly be recycled either. Reportedly, there is only one technology developed by DonNIIchermet, which can be applied for this purpose. It provides for a direct utilisation of ash residues for alloying of steel with vanadium in ladle or furnace. The degree of extraction of vanadium in this case is 70–90 % [22, 23].

The E.O. Paton Electric Welding Institute, based on its positive experience in the field of electroslag cladding (ESC) of dies, rolling mill rolls and other parts using non-compacted additive materials [24, 25], developed and tested under pilot production conditions the electroslag technologies for processing of the above wastes. These technologies solve not only technical-economic but also environmental problems. The proposed electroslag technologies for recycling of metal-containing industrial wastes are based on two types of designs of non-consumable electrodes and forming fixtures (moulds).

Non-consumable electrodes of the first type are composite electrodes having a water-cooled metal body and graphite tip. The fixed copper water-cooled mould used in this case, having a rectangular or cylindrical shape, allows cladding on a horizontal plane or production of flat ingots.

Non-consumable electrodes of the second type are sectional structures that combine functions of an electrode and forming shoe, the so-called current-conducting mould (CCM). Ingots can be produced or cladding of rather long parts can be performed using CCM [24].

Application of the first type of non-consumable electrodes and the fixed mould is provided for in the following of the developed technologies: ESC of press-forging fixture, where tool steel chips are used as an

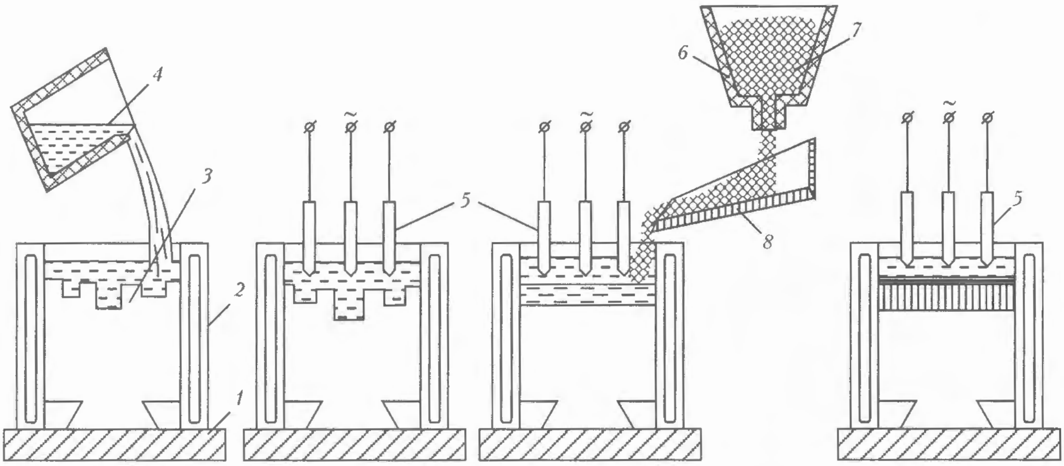

Figure 1. ESC of dies using tool steel chips: *1* — hearth; *2* — fixed mould; *3* — worn out die; *4* — ladle with liquid slag; *5* — non-consumable electrodes; *6* — proportioner; *7* — chips; *8* — spout

additive; alloying of deposited metal with vanadium through feeding the heat power plant ash residues containing vanadium to the slag pool in ESC; remelting of slime wastes resulting from grinding of high-speed steel tools to produce alloying elements for the flux-cored wire charge.

In ESC of dies, where the tool steel chips are used as an additive, first the surface of a solid billet, i.e. forging of carbon steel or worn out die, to be clad is heated using non-consumable electrodes. After incipient melting of the surface, the die steel chips, which form a deposited layer in melting, are fed to the slag pool (Figure 1). Upon melting of the required amount of the chips, the power of electroslag heating is decreased to ensure oriented solidification of deposited metal under the layer of molten slag. Particles of the chips act, at the same time, as macrocoolers to control the process of solidification of the deposited metal. Controlled oriented solidification of metal in ESC and processing it with slag during melting provide the deposited metal, the properties of which are at a level of wrought metal or sometimes even higher.

Tests of the dies clad by the above technology, conducted using industrial hammers with dropping parts 1, 3 and 5 t in weight and presses with a force of 25, 63 and 120 MN, showed that they were 1.5–3 times stronger than forged dies of the same steel grade.

Comparatively simple and reliable equipment (unit OB-2213 with electromagnetic proportioner DPE-02) was developed to implement this technology. This equipment is applied at the Joint Stock Company «Rostselmash» (Russia) and Open Joint Stock Company «Tokmaksky KShZ» (Ukraine) to recondition and manufacture dies of different weight. The technology is protected by patents of Ukraine, Russia and Belarus.

ESC of dies with chips of steel 5KhNM was performed by additionally alloying deposited metal with vanadium reduced from the heat power plant ash residues fed to slag. Operation of the dies restored by ESC with chips of steel 5KhNM indicated that additional alloying of deposited metal with vanadium (0.10–0.15 wt.%) increased their strength by 30 %.

This equipment is applied also for remelting of slime wastes formed in grinding of high-speed steel tools. Given that the range of grades of high-speed steels used currently is relatively narrow, and is often limited to steel R6M5, producing alloying elements of approximately the same chemical composition involves no serious difficulties. Non-metallic components of abrasive wheels present in the wastes remelted do not violate stability of the electroslag process and make unnecessary partial or full replacement of slag during melting.

The resulting alloying elements are used in charge of surfacing and arc metallising flux-cored wire. Coatings on parts of the type of shafts, produced by arc metallising using flux-cored wire, one component of the charge of which consists of alloying elements made by the above technology, are characterised by high wear resistance and adhesion strength.

Flow diagram of remelting of chips using non-consumable electrodes of the second type is shown in Figure 2. CCM *2* is installed on hearth *1*. Slag melted in a separate crucible («liquid» start) is pored into the mould. In some cases it is possible to melt flux using an additional electrode inside CCM («solid» start). After the slag closes all isolated sections of CCM, the current flowing through slag pool *3* provides a sufficient amount of heat released in it to maintain the electroslag process. Metal pool *4* is formed from melting the chips in the slag. The chips are fed to the slag pool using different types of proportioners *5*. The simplest solution to this problem is the use of vibrating proportioners, as in the majority

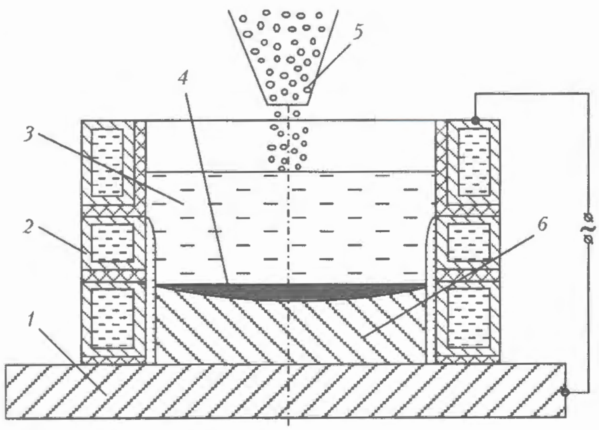

Figure 2. Flow diagram of remelting of wastes (high-speed steel chips) in CCM (see designations in the text)

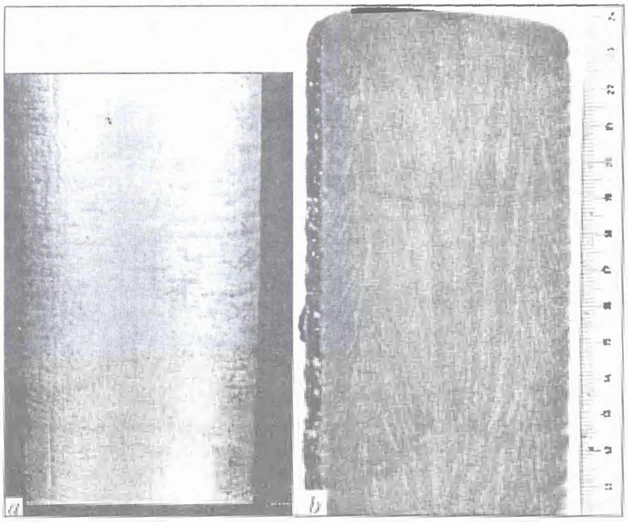

Figure 3. Appearance (*a*) and longitudinal macrosection (*b*) of R6M5 steel ingot produced by remelting of high-speed steel chips

of cases no very precise proportioning of an additive remelted is required. Portions of liquid metal gradually solidify to form ingot *6*, like in conventional electroslag remelting (ESR).

Unit OB-2379 equipped with a drum-type proportioner and a set of CCM 40–120 mm in diameter was developed and manufactured for ESR of chips. Ingots produced by the unit are up to 0.7 m long, which is provided by their withdrawal from fixed CCM. The slag pool is induced by the «solid» start, which is the simplest method of melting flux AN-75. Flux of this grade provides ingots with a smooth surface, having almost no skull. Figure 3 shows the surface of the R6M5 steel ingot after sand-blasting, as well as the longitudinal macrosection of the ingot. Increasing the mould diameter makes the remelting process simpler, raises productivity of the process and decreases its power intensity. For example, the process productivity achieved in remelting of the R6M5 steel chips using CCM with a diameter of 210 mm is 200 kg/h, the specific power consumption being 1000–1200 kW·h/t.

In the majority of cases, to produce ingots with chemical composition corresponding to that of high-speed steel, the chips prior to remelting should be cleaned from emulsion with hot solution of soda ash or baked in furnace at a temperature of about 300 °C.

The tools made from the remelted high-speed steel chips were proved to be not inferior in strength to the tools made from wrought high-speed steel.

CCM was used also to remelt wastes in the form of copper wire scraps of a different length, which were preliminarily transformed into chops 5–7 mm long. Quality of the resulting copper ingots depends upon the uniformity of feeding of the chops and the amount of insulation coatings left in them. The ingots produced can be utilised to manufacture different copper parts.

CONCLUSIONS

1. The developed technologies allow chips of die and high-speed steels to be utilised as an additive for cladding of tools used for hot deformation of metal. Such chips can be remelted into ingots, from which different metal-cutting tools are made.

2. Ash residues containing oxides of vanadium and other metals, which are formed in combustion of fuel oil at heat power plants, can be utilised in electroslag processes for alloying remelted or deposited metal. Alloying is provided by reduction of metals from their oxides in the slag pool.

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PROPERTIES OF DEPOSITED METAL USED FOR METALLURGICAL TOOL HARDENING

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Problems associated with rational alloying of deposited metal of the Ni–Cr–Mo–Nb–C system to provide hardening of metallurgical tool operating at temperatures up to 900 °C are considered. The effect of molybdenum and niobium content on high-temperature hardness, susceptibility to hot cracking and heat resistance of deposited metal has been studied.

Keywords: *arc hardfacing, high-alloy deposited metal, high-temperature strength, heat resistance, hot cracks, flux-cored wire*

Alloys with a high resistance to structure homogenization are preferable for the wear-resistant hardfacing of metallurgical tool subjected to a cyclic temperature-force effect (CTFE) at maximum temperature up to 900 °C at the site of contact with a workpiece [1]. In this respect, the deposited metal on the base of nickel is promising, in which the coefficients of diffusion of atoms of a base, alloying and impurity elements at 900 °C are low and equal to about $1\cdot10^{-13}$ cm^2/s [2]. Alloys with a high energy of activation of diffusion, that is provided by their alloying by chromium, molybdenum and carbon, having a limited solubility in nickel, are especially effective [3, 4]. However, the resistance of this deposited metal to hot crack formation is low, as was shown by the experimental checking. The hot cracks are caused by a high content of eutectic carbides $Me_{23}C_6$, Mo_2C and large carbides of Me_7C_3 type, precipitated along the boundaries. The technological properties of carbon deposited metal can be improved by its alloying using the more active carbide formers, as compared with chromium and molybdenum, for example with niobium [5]. In this connection, the effect of ratio of alloying elements on high-temperature hardness, rate of oxidation, susceptibility to hot cracking of metal of Ni–Cr–Mo–C–Nb system, deposited by flux-cored wires using argon arc and electroslag methods, was studied.

The high-temperature hardness of the deposited metal was measured in unit TSh-2 using a hard-alloy 5 mm diameter ball at 7.35 kN load and 10 s holding. Oxidation rate was determined after holding of samples in electric furnace at 900 °C temperature for 2 h. Susceptibility to hot cracking was evaluated by their amount in metal of a five-layer bead in argon arc hardfacing.

The flux-cored wire sheath made from nickel strip NP-2 was manufactured two-layered with displacement of butts relative each other to improve the technological reliability of wire at high (up to 60 %) coefficient of its filling with charge [6]. As the charge materials, the metal powders of nickel, molybdenum, chromium, and also graphite and ferroniobium FN-1, containing impurities of titanium, aluminium and tantalum, which, being active deoxidizers, improve the quality of the deposited metal, were used. Metallographic examinations of the deposited metal were carried out using known methods, the X-ray spectral analysis was made in the Cameca unit MS-46 with microprobe diameter of 1 μm.

Range of alloying of experimental deposits is as follows, wt.%: 1.58–2.84C; 19.6–24.5Cr; 2.85–9.5Mo; 0.65–2.86Nb. Content of other elements, wt.%: 0.60–0.85Si; 3–5Fe; 0.4–0.6Mn; 0.012–0.015S; 0.004–0.005P; Ni — base.

The selected range of alloying of deposited metal is stipulated by service and technological requirements to it. The amount of oxygen was defined from the feasibility of providing the atomic ratio of carbide-forming elements and carbon of not less than stoichiometric ratio, at which special carbides MoC and NbC are formed, and also from the experimental data [7]. The ranges of alloying by chromium are determined coming from obtaining the maximum heat resistance of the metal. It was established experimentally, that increase in chromium content up to 30 wt.% and more does not lead to the significant increase in heat resistance of alloys of the alloying systems considered, and is manifested only in the increment of their hardness at room temperature that deteriorates the metal workability after hardfacing. The upper limits of alloying with molybdenum and niobium are stipulated by the presence of a high content (up to 60 vol.%) of hardening phases in the deposited metal, that influences greatly the ductility and its resistance to cracking during hardfacing. With decrease in molybdenum and niobium content to the lower limit, the high-temperature hardness of alloys is abruptly decreased.

It was established by investigations (Figure 1) that at ratio of molybdenum and carbon in the ranges of 0.18–0.25 at.% , contained in the deposited metal, a sufficiently high level of its properties is attained for operation under the CTFE conditions at temperature up to 900 °C in the zone of contact with a metal being subjected to working. Structure of this alloy consists of a solid solution on the base of nickel, carbides NbC and MoC, and also two types of carbide eutectics with microhardness *HV* 350 and *HV* 400.

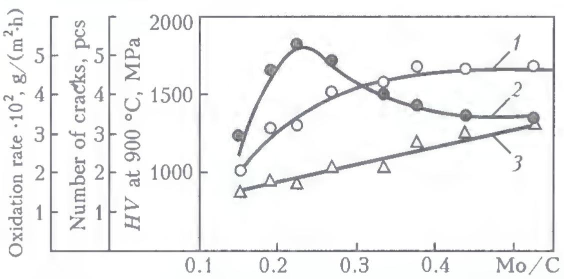

Figure 1. Effect molybdenum/carbon ratio on properties of deposited metal with 22–24 wt.% Cr and 1.4–1.8 wt.% Nb: *1* — susceptibility to hot crack formation; *2* — high-temperature microhardness; *3* — oxidation rate

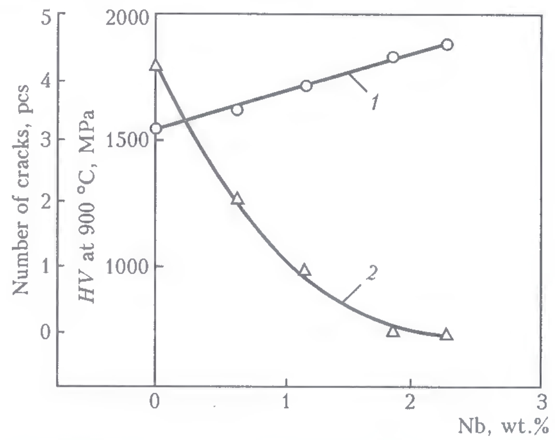

Figure 2. Effect of niobium on high-temperature hardness (*1*) and resistance of deposited metal against hot crack formation (*2*) at the following content of elements, wt.%: 22–24Cr; 4.5–4.8Mo; 2.0–2.5C

It is shown that a good resistance of deposited metal against the hot crack formation and maximum high-temperature hardness are provided at niobium content of 1.8–2.5 wt.% (Figure 2). This effect of niobium is associated with its strong liquation caused by carbide formation (Figure 3), that leads to grain refining, crushing of carbide eutectics and ensures the comparatively uniform distribution of silicon in eutectics and metal grains (Figure 4). Molybdenum is also distributed uniformly that is explained by its strong chemical bond with nickel and it is one of causes of increased heat resistance of these alloys as compared with the deposited metal containing tungsten instead of molybdenum.

Increase in ratio of atoms of molybdenum and carbon above 0.25 leads to the formation of Mo_2C carbides in metal, which are less stable thermodynamically than MoC. The high-temperature hardness and heat resistance of deposited metal in this case is decreased, and susceptibility to hot crack formation is increased. Decrease in heat resistance can be explained by the formation of an easily-sublimating oxide of molybdenum at the metal surface. The crack formation is increased as a result of growth of a volume content of eutectic carbides.

Under the model condition of CTFE (150–900 °C change in temperature; force — 300 MPa; time of contact without allowance for pauses — 2 s; number of test cycles — 500) the metal of the following chemical composition was tested, wt.%: 2.55C; 4.5Mo; 22.5Cr; 2.1Nb; Ni — the base. After tests the

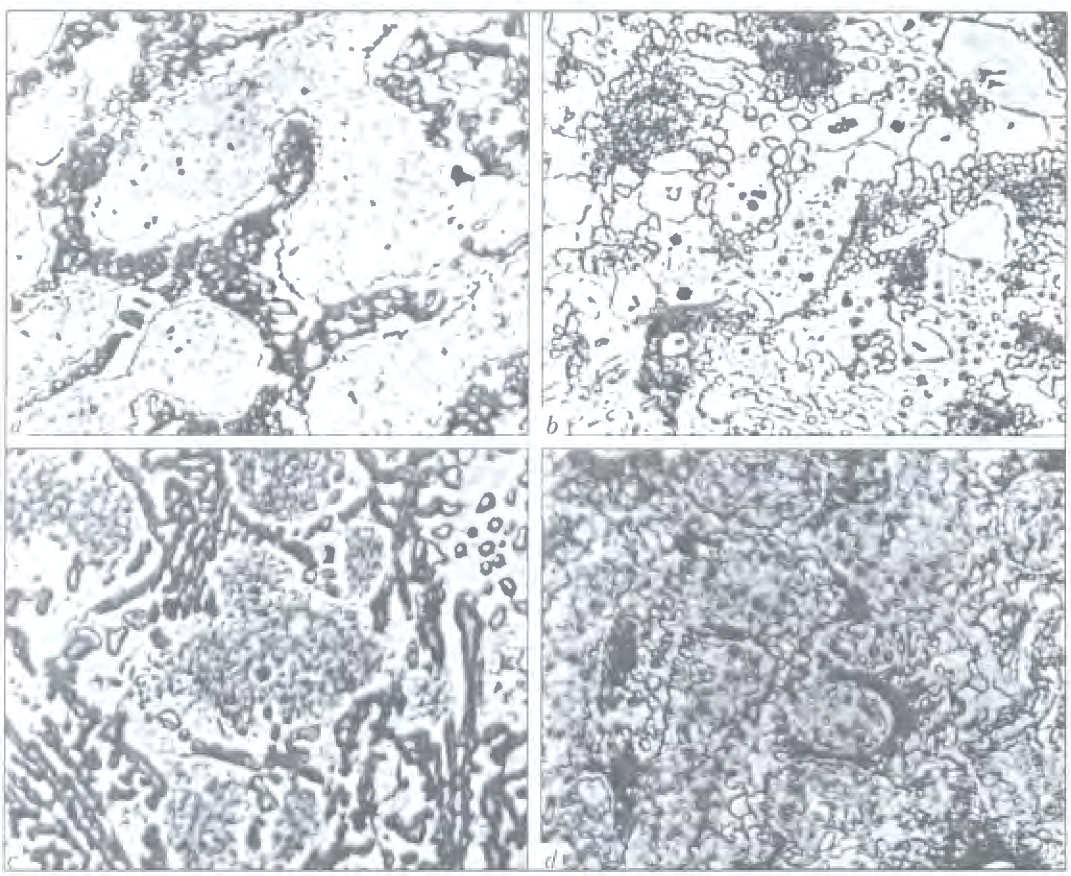

Figure 3. Microstructure of deposited metal without niobium (*a*) and with niobium content, respectively, 1.1, 1.82 and 2.35 wt.% (*b–d*) (×600)

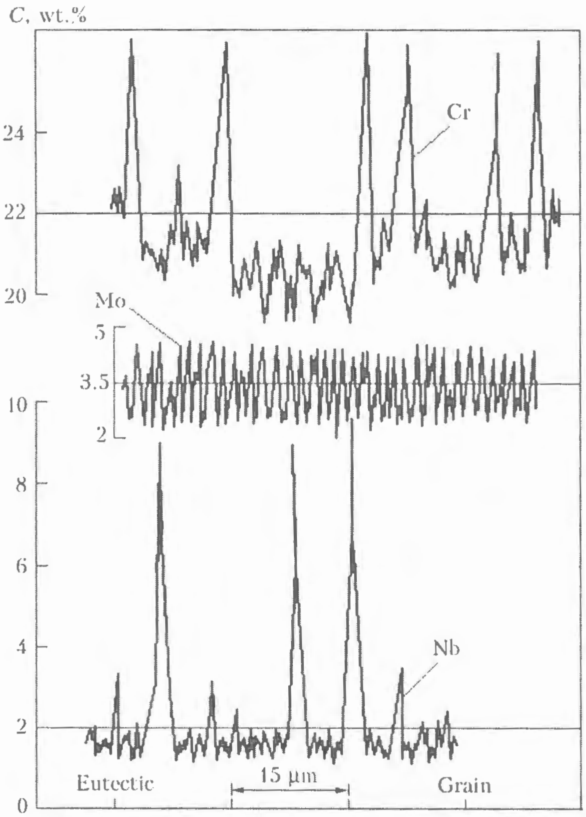

Figure 4. Distribution, *C*, of alloying elements in deposited metal

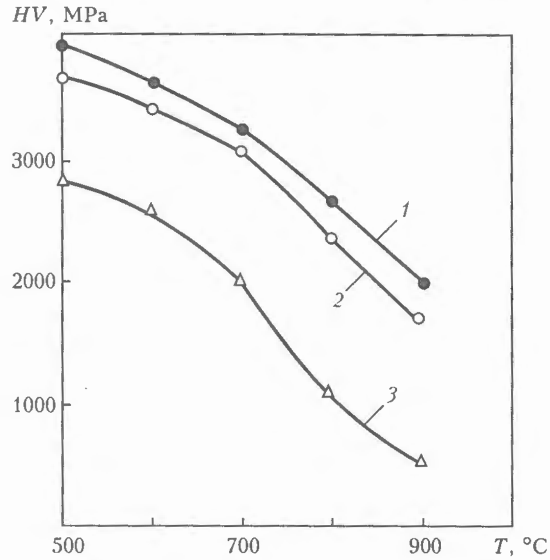

Figure 5. Effect of test temperature on high-temperature microhardness of deposited metal: *1* — Hastelloy C; *2* — 250Kh22N66M4B2; *3* — 30Kh2V8FS

noticeable change in structure and properties of the deposited metal was not observed. Hardness was negligibly increased (by *HV* 50–60) as compared with initial hardness (*HV* 400–420) due to decomposition of solid solution and formation of secondary carbides (Figure 5).

The developed composition of flux-cored [8] (250Kh22N66M4B2 type of deposited metal) has passed the industrial tests in hardfacing the cutting edges of knives for cutting the hot metal, lips of mandrels of pipe-rolling unit, press dies for hot working of metals. The wear-resistance of deposited tool was 2–2.5 times increased as compared with industrial types of deposited metal of Fe–C–Cr –W–Mo system.

CONCLUSIONS

1. Deposited metal of Ni–Cr–Mo–Ni–C system of alloying was suggested to harden the metallurgical tool operating at temperatures of up to 900 °C. It was established that to provide high technological and service properties of the deposited metal of this type, the ratio of molybdenum and carbon content in it should be in the 0.18–0.25 ranges.

2. Increase in Mo/C ratio by more than 0.25, as well as the increase in molybdenum content in deposited metal above 5.0–5.5 wt.% does not promote the growth in its high-temperature hardness, stipulates the reduction in heat-resistance and resistance to hot crack formation.

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NEW EQUIPMENT FOR HARD-FACING OF BLAST FURNACE BELLS AND HOPPERS

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Design peculiarities and technical capabilities of the device developed for wide-layer hard-facing of metallurgical equipment are briefly described.

Keywords: *wide-layer arc hard-facing, bells and hoppers, welding device, technological capabilities, specifications*

Hard-facing of blast furnace bells and hoppers is performed using unique machines U-50 and U-75 (Figure 1). The machines comprise manipulators with a load-carrying capacity of 50 or 75 t and a mobile column with the welding device A-1640 mounted on it. The column causes electrode weaving with amplitude of 50–400 mm, while the manipulators provide movement of a workpiece to a welding deposition pitch. The machines are designed with a possibility of performing circumferential deposition and welding of large-size parts by the open-arc or submerged-arc method using flux-cored or all-drawn wires. Many years of utilisation of these machines at metallurgical and machine-building plants of the CIS countries proved their high reliability and good performance. However, the welding devices and electric circuits have become obsolete, both physically and morally, and require renovation.

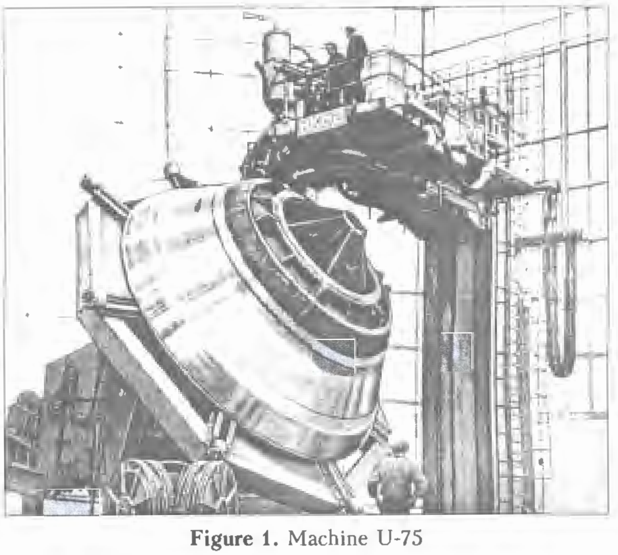

Figure 1. Machine U-75

The E.O. Paton Electric Welding Institute in collaboration with the Limited Liability Company PLAN-T developed a new device A-1812 and electric circuit SU-300 to complete machines U-50, U-75 and U-125.

Device A-1812 (Figures 2 and 3) has the following peculiar properties:

- feeding mechanisms are equipped with asynchronous motor drives, which provide a more uniform feed of the electrode material;
- travel of the beam is increased by 150 mm, while the A-1812M-01 modification is designed with a possibility of moving the device out to another 750 mm using slides, which is important for hard-facing of long internal surfaces;

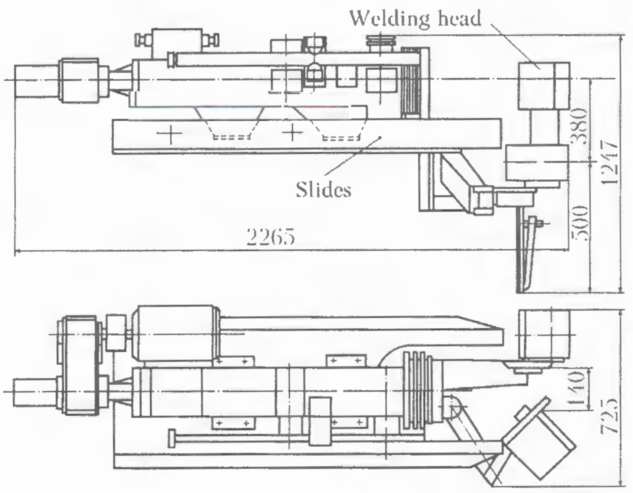

Figure 2. Schematic of device A-1812

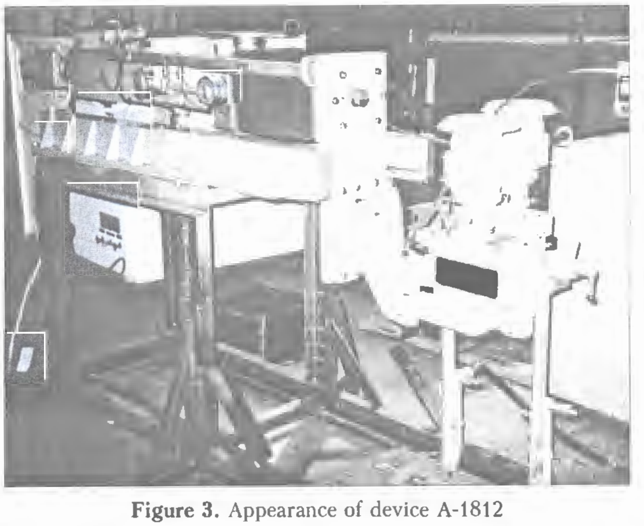

Figure 3. Appearance of device A-1812

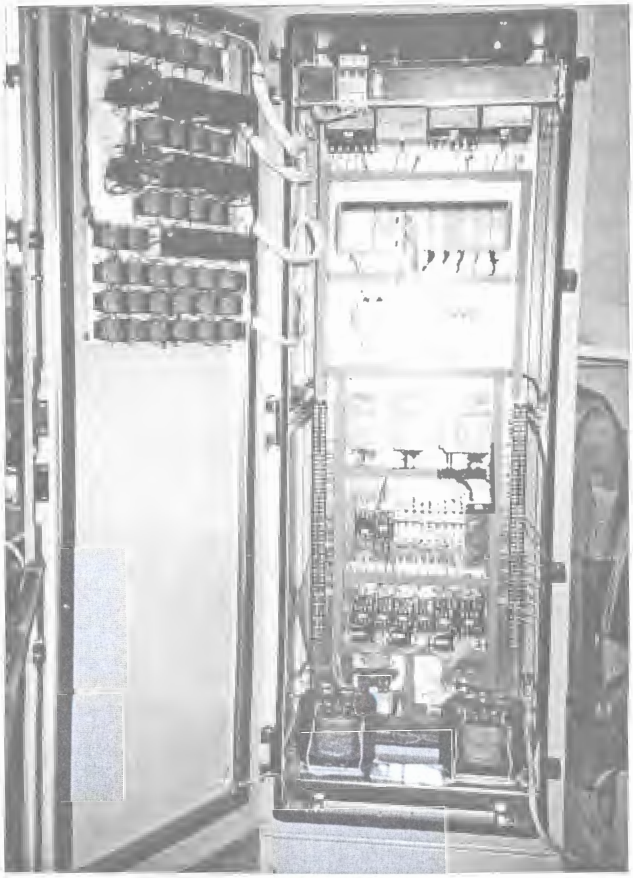

Figure 4. Control cabinet SU-300

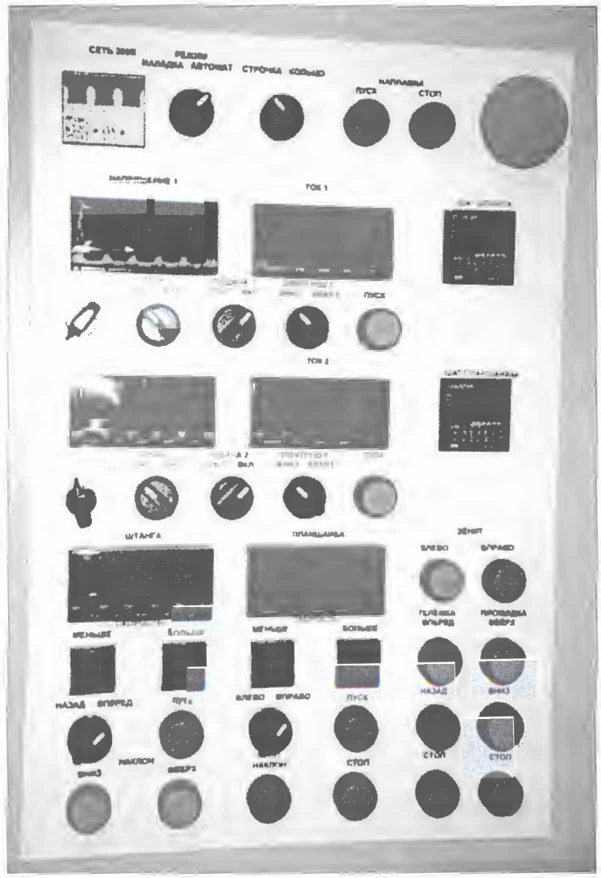

Figure 5. Control panel SU-300

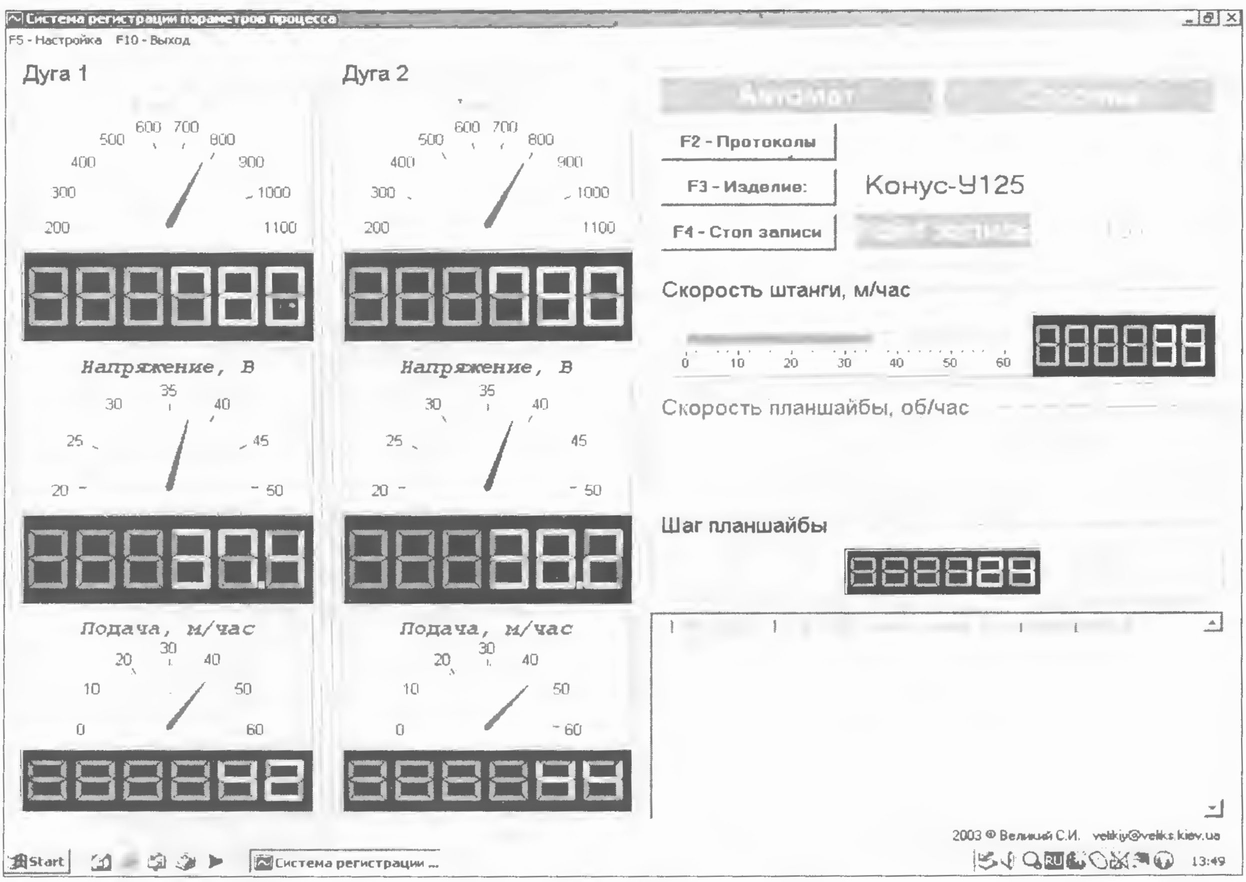

Figure 6. Logging system interface

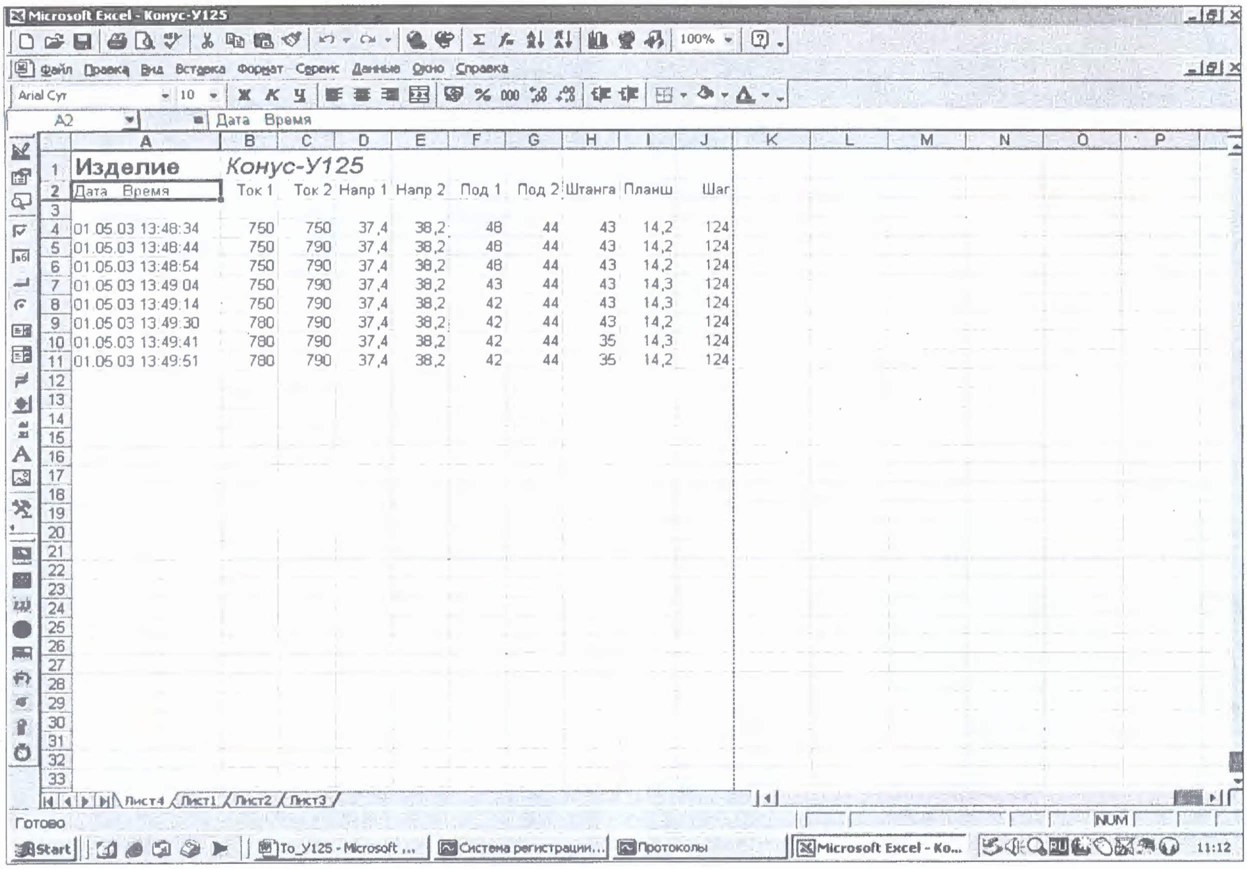

Figure 7. Logging of process parameters

• brackets of feeding mechanisms allow welding deposition to be performed simultaneously with the two arcs, either successively or in parallel;

• the device is fitted with an indexer, which provides increase in accuracy of displacement of a workpiece by a deposition pitch and allows adjustment of the pitch directly from the control panel;

• device A-1812 is completed with the microprocessor-based control system SU-300, which controls actuating mechanisms of the entire hard-facing machine based on logical processing of the data received from different sensors.

The distinctive feature of the new electric circuit is a total refusal from DC motors. They are replaced by the asynchronous motors providing a smooth adjustment of the rotation speed. The circuit enables display of all controlled parameters on the monitor screen with a possibility of real-time logging of the deposition process conditions, which allows certification of the manufacture of hard-faced parts.

Physically, the new electric circuit consists of two cabinets, which serve as control panels (Figures 4 and 5) and, at the same time, as personal computers. One of the cabinets with the master control panel is located on a work station, where a workpiece to be hard-faced is fixed (using clamping jaws). The latter allows a 2–3 times reduction of the quantity of cable lines.

Readings of all instruments used to monitor the process parameters are logged on to the computer (Figure 6), and the deposition process conditions are recorded in real time (Figure 7). The data thus obtained can be stored on a hard disk or other information carrier. All this information can be viewed on the monitor screen or printed out.

Specifications of device A-1812

Cross section of flux-cored strip, mm	$\leq 18 \times 4$
Electrode wire diameter, mm	≤ 3–5
Arc voltage, V	28–50
Electrode feed speed, m/h	20–70
Speed of reciprocal movement of electrodes, m/h	10–70
Stroke of reciprocal movement of electrodes, mm	≤ 550
Horizontal travel of welding device along slides, mm	≤ 750
Duty cycle, %	100

The new device and control system were manufactured and validated. The new equipment has been commercially applied to advantage since January 2004 at the «Krivorozhstal» plant.

Assembly and industrial tests of the new equipment were completed in September 2004 at the Open Joint Stock Company «Azovmash». Application of this equipment allowed widening of its technological capabilities, enhancement of its reliability and certification of the hard-facing process, which basically solved the problem of improvement of the quality of treated parts.