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IMPROVEMENT OF EFFICIENCY IN ESR METAL PRODUCTION

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The state-of-the-art of ESR metal production is analyzed and prospects for the development of electroslag technologies and equipment, which are based on the electroslag process, are considered. Assortment of ingots, billets and semi-finished products is offered whose production can be realized effectively in the industrial conditions after modification (or reconstruction) of the electroslag equipment and implementation of new electroslag technologies.

Key words: *electroslag remelting, electroslag surfacing, ESR with liquid metal, ESR furnace, chamber ESR furnace, ESR ingots, bimetallic ESR ingots*

Analysis of the tendencies in the development of main heavy industry branches, primarily the modern machine building, shows that without using the newest technologies of special metallurgy in production of advanced structural materials and different half-finished products of high-alloy steels and alloys with preset properties it is impossible nowadays to produce metallurgical and machine-building products competitive at the world market.

Therefore, it is extremely important to determine a right direction for production of the ESR metal in future. The urgency of this task for Ukraine is specified by the sufficiently large number of ESR production capacities in the country, which were created as early as in the 60–80s both in metallurgy («Dneprospetsstal», «Azovstal») and machine-building (Novo-Kramatorsk Machine-building Plant) mainly to meet the demands of the defense industry, power engineering, ship-building and transport. Today these capacities are practically unloaded.

Abroad in such countries as USA, Germany, Great Britain, Austria, Italy, Japan and others many ESR shops that were oriented for production of high-speed steel, superalloys (in technological chain VIM + ESR + VAR) and large ingots for power machine-building continue successfully and profitably operating as well as new ESR furnaces are being constructed [1–3].

The world leading firms presented at the 14th International Forgemasters Meeting [4–5] organized by the German Institute of Ferrous Metallurgy their report testifying to the effective application of the ESR for production of large ingots for rotor forgings and discs for powerful steam and gas turbines.

The German firm «Saarschmiede GmbH» together with the American firm «GE Power System» forged from the ESR ingot 2300 mm in diameter and 134 t mass a rotor shaft billet of 1845 mm in diameter for single-block power units of low and intermediate pressure of the steam turbine. Subsequently the billet was

subjected to differential hardening. Despite of the gigantic size of the rotor billet the forging metal was distinguished by a required combination of high strength ($\sigma_{0.2} = 700$ MPa) and impact strength (brittle transformation temperature is -53 °C) [6].

The report [7] presented by Japan firms «Kobe Steel» and «Mitsubishi Heavy Industries» informs that a 74-ton electroslag ingot of high-chromium (up to 12 % Cr) steel of the new generation they developed (with additives of molybdenum, tungsten, cobalt, niobium, vanadium and boron) was forged to produce a billet for the steam turbine rotor operating under 630 °C. Quality control of the produced rotor billet showed that resistance of this ESR steel under creep test is 1.5 times higher than in the steels used before.

Germany in the last decade considerably increased ESR production capacities. The firm «Gröditzer Stahlwerke» put into operation in September 1996 a new doubled ESR unit [8], which allows remelting the consumable electrodes 300–900 mm in diameter, up to 4000 mm long and up to 12 t mass into ingots 500–1650 mm in diameter, length in the range of 2.5 up to 5 m and mass within 9–84 t. The unit is equipped with a device for ingot withdrawal and electrode replacement. It has two movable gantries where the systems for feed and replacement of the electrodes are located. These systems handle two remelting stands. It is noteworthy that the «Gröditzer Stahlwerke» before putting the ESR furnace into operation had already operated two steel-melting facilities of 100 thous. tons of annual capacity. Two arc furnaces (30 and 5 t), ladle furnace, vacuum-oxygen decarburization unit and main steel-teeming ladles with slide gate and bottom blowing are also operating there.

The firm «Krupp VDM GmbH» invested approximately 15 mln DM into construction of the ESR and VAR furnaces about 4 thous. tons of remelting capacity each in the territory of the plant in town Unna where the melting and casting operations of the firm are concentrated [9]. Putting into operation of new furnaces will allow producing structural parts of high alloys demanded by the aviation industry and materials with specific magnetic characteristics for elec-



tronic industry. This will open new markets to the firm.

Along with electroslag remelting in conventional furnaces there is an increasing demand for chamber ESR furnaces in the controlled atmosphere [10]. Under remelting of steels and alloys alloyed with lightly oxidized elements it is necessary to protect the slag pool and electrode from ambient air with inert gas. It is noteworthy that the first in the world chamber ESR furnace U-102 designed for electroslag remelting of precision steels and alloys was especially developed at the E.O. Paton Electric Welding Institute of the National Academy of Sciences of Ukraine and put into operation as early as in the 60s [11]. The furnace U-102 is equipped with an air-tight chamber with a removable cap for inserting a consumable electrode and provides the ESR process under full air-tightness inside the mould and around consumable electrode. The furnace works by a scheme of filling the stationary mould and produces ingots 150 mm in diameter and 1000 mm long in the controlled gas atmosphere under normal pressure. Due to a number of reasons the chamber furnaces did not gain acceptance in our country. However, new promising application areas have appeared for ESR chamber furnaces in the last decade. This has been necessitated by remelting of steels and alloys alloyed with lightly oxidized elements and by peculiarities in using metallic calcium and rare earth metals under ESR [12]. Chamber furnaces are also required for industrial implementation of the arc-slag remelting developed at the PWI especially under alloying of metals with nitrogen from the gas phase and producing high-nitrogen steels and alloys also under pressure fully excluding the use of expensive nitrogen-containing additives (nitrided ferroalloys, silicon nitride and others) from the technological process [13, 14].

The known firm «ALD Vacuum Technology» with its branches in Germany and USA proposed the ESR furnace design with a protection atmosphere, which can operate under pressure [15]. An ingot 800 mm in diameter with practically similar content of CaO, FeO, MnO, Cr₂O₃ in its lower and upper part but with varying content of SiO₂ within 0.6 to 2.8 % was produced in this furnace. This variation lies within 7.2–11.2 % for conventional ESR. Besides, aluminium, titanium and silicon burnt out to a lesser degree.

The German firm «VSG Energie- und Schmiedtechnik GmbH» applied the ESR technology under pressure in the controlled atmosphere for melting large ingots (1000 mm in diameter) of high-chromium steel (0.31 % C; 0.55 % Si; 15.5 % Cr; 1.02 % Mo) with super-equilibrium nitrogen content (0.38 %). Equilibrium content of nitrogen in this steel under air pressure and temperature 1600 °C in the liquid state is only 0.04 %. It is noteworthy that the actual resistance of bearings produced of this steel proved after trails 80 times stronger than the designed resistance provided that the bearings were not brought

to destruction [16]. The same German firm together with the branches of the Swiss firm «ABB Alstom Power Technology» developed high-nitrogen (up to 0.16 % N) high-chromium (up to 12 % Cr) steel with yield strength 1100 MPa for gas turbine disc applications. Large ingots were melted in the chamber furnace by the conventional ESR technology for high-chromium steel [17].

The firm «Inteco» (Bruck, Austria) in 1997 constructed and put into operation at the firm «Acciaierie Valbruna» in Vicenza (Italy) the ESR unit performing electroslag remelting in the controlled protective gas atmosphere created inside the cap tightly placed on the short water-cooled mould [18]. Square 140 and 200 mm and round ingots 100, 140 and 160 mm in diameter, up to 4 m length and up to 6 t mass were produced by the so-called Electroslag Rapid Remelting (ESRR) patented by «Inteco». In this method a section of a consumable electrode 3–10 times exceeds a section of the ingot drawn from the mould. Quality of ESR ingots of stainless steel and special alloys melted in this furnace with the electrode exchange meets the requirements imposed on the half-finished products for power, aviation and nuclear engineering by the respective standards [19].

Quality changes in production of ingots and half-finished products of ESR metal have been observed in the last years at the plants of the Russian Federation and Ukraine.

Production of pipes for critical applications (for nuclear engineering, oil and gas complex and others) by a direct ESR ingot rolling has been mastered at the Chelyabinsk Metallurgical Works and Chelyabinsk Pipe Rolling Plant [20]. Cast consumable electrodes of 1.45–2.3 t single mass exchanged in the course of melting were used to produce ingots 460, 485, 500 and 600 mm in diameter and 2.1–3.7 t single mass in the furnaces R-951 and EShP-2.5. An increase of a single mass of ingots from 1.5 up to 3.7 t and optimization of their weighing have allowed considerably expanding a range of the metal products and raising their competitiveness, which have resulted in full utilization of production capacity of the ESR furnaces.

Moreover, the available vacuum-arc furnace DSV-4.5 and DSV-8.0 due to a decreasing consumption of the VAR metal and increasing demand for the ESR metal are operating as ESR units [21] at the Chelyabinsk Metallurgical Works. For this purpose the furnaces were equipped with the device for feed of flux and deoxidizers and a system for suction of the released gases. Moulds, bottom-places, electrode-moving mechanisms, power supply sources were used without changes.

Increasingly more attention at the plant «Dneprospetsstal» and metallurgical works «Azovstal» is paid lately to seeking new solutions on creation of resource- and power-saving technologies, increasing a range of the produced ESR metal and decreasing its production cost [22–25]. Putting into



operation of the first industrial ESR furnace R-909 at the «Dneprospetsstal» in May 1958 initiated a rapid development of electroslag remelting. For this period more than 200 grades of the ESR steel have been developed at the plant with the ESR ingot production achieving 100 thous. t a year. Now about 1.5 thous. t of ESR steel is produced monthly, 20–25 % of the metal being exported abroad. A range of ESR steels and alloys produced by the plant counts 60 grades (chromium and chromium-nickel stainless, bearing, structural and tool alloyed, high-speed and other steels).

With the aim of further expanding of the production volumes on the available capacities the plant has developed new ESR products including chisel steels, heat-resistant alloy EI15U, tool steel Kh12F [22]. For decreasing a cost of electroslag metal a new technological scheme has been developed. According to this scheme half-finished strips 360×170 mm in section rolled from 4.36 t ingots on the blooming mill of the plant are used as consumable electrodes for ESR. Previously the rolled slabs obtained on cooperation basis and produced at the «Zaporozhstal» plant were used for this purpose. This allows under preserving the desired quality of the high-strength plate steel a decrease of its cost by 15 % and a reduction of the metal process stage by half a month.

A technology for continuous casting of 370×370 mm consumable electrodes for production of ESR ingots in the 565×565-mm moulds has been improved. This has permitted increasing utilization efficiency coefficient of the initial metal and yield of the rolled stock of steel 20KhN3ASh from 67.7 up to 73.5 % with the steel quality meeting the standard. Considerable effect was produced at the plant by a repeated utilization of the spent fluxes (up to 40 %) with regard to their regeneration, thus allowing for 10–15 % decrease of their consumption against the earlier indices [23].

With the aim of saving material and power resources at the «Azovstal» the ESR furnaces 20 VG along with production of large plate 580×1160 and 580×1900 mm ingots 14–23 t mass for the first time started producing 800×1160 and 580×800 mm billets. The consumable electrodes used for this remelting were produced by a longitudinal cutting of continuously cut 300×1850 mm slabs of metal produced in plant own oxygen converter [24].

It should be noted that it is impossible to increase the competitiveness of the ESR metal without a decrease of its cost mostly determined by the expenses on production of consumable electrodes and the ESR metal process itself. Production costs can be essentially decreased by using the spent long-dimensional products as consumable electrodes.

A technology for ESR utilization of different metallurgical wastes from the plant own production (spent continuous casting rolls, worn-out panels of copper moulds, scrapped rails, plate cuts) has been developed at the «Azovstal». Under assembling of

consumable electrodes different waste parts are joint by welding. As a result of these efforts the ESR efficiency has been increased both economically and ecologically by reduction of the consumable electrode production stages. No crucial changes in the ESR furnace design were needed [25]. Similarly the spent cast iron rolls 250 mm in diameter butt joint in pairs were used as consumable electrodes by authors of [26]. In this case the electric power consumption under remelting in the widened mould 250 mm in diameter by a scheme of drawing of the cast iron billets intended for flour-milling rolls did not exceed 1000 kW/t while the ingot cladding velocity achieved 400 kg/h.

Modification (reconstruction) of the existing ESR furnaces along with creation of new ones is now one of the possible and promising ways for increasing efficiency of electroslag technologies for producing solid and hollow billets (ingots) of different steels and alloys. In modern conditions of market economy the targets set for this modification are rather tough: decrease of the cost of production or achieving such quality of the product, which could fully meet new increased demands of the consumers, be competitive and provide minimal period for recoupment of the increased expenses.

However, a level of main constructive, technical and technological solutions provided in the modification project that should be at least on par with the achieved world levels is undoubtedly the most important and decisive conditions determining success and terms for fulfilling the set task.

In this connection a high quality level of the automation of modern ESR furnaces appeared in the last years abroad should be especially emphasized. First of all new ESR units are fully computerized, equipped with perfect systems of automatic control, in particular with system for automatic weighing of consumable electrode, batch feed of flux and deoxidizers, pickups of the metal pool level in the mould and others.

Electric power costs are one of the most important expenditure items in the ESR metal process. A specific consumption of electric power varies considerably from melt to melt. Energy consumption mainly depends on the furnace design, first of all on the design of short (power) mains. Application of the optimal design of the short furnace mains may considerably save the electric power (by 20–50 %). Modern one-phase ESR furnaces have compensated short mains with a high degree of coaxiality of electrodes and current leads.

Melting of ESR ingots with increased space factor (up to 0.7–0.8 instead of 0.4–0.5 conventional for us), use of smaller slag volumes than it is adopted in this country, increase of the slag electric resistance, remelting of the electrode with the melting velocity approaching the upper permissible limit (with a flat electrode face), automatic control of technological melting process not only by electric parameters and mathematical models but also by other direct measurements, for example, electrode mass — this is the



list of main features characterizing remelting technology in modern furnaces.

One of the original solutions of the last years found in Ukraine consists in the new process of electroslag remelting of ingots in current-supply mould by the two-circuit scheme [27, 28] when the electric power to current-supply mould and consumable electrode is supplied in two separated circuits from separate power sources. As a result electric power supplied to the non-consumable electrode is mainly consumed for heating of peripheral (lateral) part of the slag pool providing a good formation of lateral surface of the ESR ingot drawn from the mould. Consumable metallic electrode, which is usually located in the center of the slag pool, is melting mainly by the heat released in the slag pool from the electric current supplied directly to the electrode.

Therefore, the two-circuit ESR scheme qualitatively changes not only the melting conditions of the consumable electrode but also a shape (geometrical parameters) of the metallic pool, thus changing a nature of the ingot solidification. In this case in contrast to the classical ESR scheme there appears a possibility to considerably weaken a dependence of the electrode melting velocity on the supplied electric power allowing for a wider variation control of the metallic pool depth to form the regulated structure of the ingot and to provide a preset level of physical-mechanical properties irrespective of the process productivity. The use of the two-circuit ESR scheme proves very promising for creation of new technological processes where application of the classical scheme is limited, for example, for production of large ESR ingots of superalloys exposed to the so-called ghost.

Analysis of the latest achievements in the sphere of electroslag technologies in Ukraine and in the world based on numerous publications in the home and foreign literature as well as on discussions during many international symposiums including the recent International Memorial Symposium to commemorate Prof. B.I. Medovar (MMS-2001) [29] testifies to the fact that a new generation of electroslag technologies and equipment, which denies the use of consumable electrodes and adopts the use of the liquid metal under ESR (ESR LM) is rather perspective [30].

It should be recognized that at present the electroslag remelting as a method of metal refining from oxygen, nitrogen, sulfur and for producing steel free from non-metallic inclusions has lost its leading position. A level of mass steel production allows today technological processes providing production of especially pure liquid metal in the necessary quantities and rather cheap. However, even the pure liquid metal does not always provide in industrial conditions a production of ingots with guaranteed high quality of microstructure, chemical and physical homogeneity at a level of the ESR metal. None of the known processes of the liquid metal casting is comparable to ESR in these important indices.

A role of ESR process among the methods of special electric metallurgy is now essentially expanding due to the appearance of the ESR LM. First of all a creation of continuous electroslag remelting with a liquid metal is implied (CESR LM).

A possibility to produce ingots of not only square and round cross section but also hollow ingots of the materials of a wide composition variety (stainless, structural and high-speed steels, heat-resistant nickel alloys and others) and to decrease the ESR cost is a doubtless advantage of this method as compared to the previously mentioned rapid ESR (ESRR).

An essential advantage of the ESR LM process is a possibility to apply it for high-production cladding under manufacturing of composite and bimetallic finished and semi-finished products (rolls, plates, section rolled stock, tubes and others).

In the mid of the 90s there appeared numerous publications [31–34] devoted to different aspects of the use and production of bimetal by the methods on the basis of the ESR process. It is noted in these works that production of corrosion-resistant bimetallic billets by the ESR method provides the highest strength of the joint of the cladding and main layers ($\tau_{\text{shear}} > 350 \text{ MPa}$) and maximal continuity of the layer cohesion in the area of the plate as compared to other methods of bimetal production. This allows producing bimetallic rolled stock of any thickness where the quality of the layer joint does not depend on the deformation degree of the initial bimetallic billet. Development of the technology for vertical electroslag surfacing (ESS) of bimetallic billets of rectangular and square sections on the available ESR furnaces, which according to [33] are applied for this practically without reconstruction of the basic elements and units is reported. It is also noted [34] that the use of the ESR process allows formation of the surfacing layer but not of the bimetallic ingot itself. Under correlation of layers 1:5 a consumption of electric power, flux and cooling water and a wear of the equipment per 1 t of the product (bimetallic ingot) is 5 times less than under ESR of the homogeneous ingot of the same mass. Initial materials for ESR of bimetallic billet and power sources are the same as under conventional ESR. These are only qualitative indices that change, in particular power consumption for 1 t of the clad bimetallic billet is 400 kW·h.

It is noteworthy that the idea to use the electroslag remelting for production of bimetallic billets was for the first time proposed as early as in the 60s of the previous century at the PWI [35]. However, this idea was not implemented at the industrial scale since the main difficulty, i.e. large and heterogeneous penetration of the basic metal under cladding of the surface layer was hard to overcome [36]. This causes a wide range of variations in the thickness of the cladding layer even in one plate. Judging from the literature this problem is still valid when the classical ESR is applied.



Appearance of the ESR LM process opens new potentialities for improving a technology for production of bimetallic billets [37–39]. The made research has shown that the application of the electroslag surfacing by the liquid metal (ESS LM) allows surfacing a clad layer with a minimal penetration depth of the basic layer both in the section and in the height of the product. In case of need it is possible to use this method for surfacing by a welding-brazing scheme, i.e. practically without guaranteed penetration of the basic layer [38, 39]. In this case the contact surface of the basic layer of the billet is treated with the melt of the synthetic slag to purify it from oxides and other impurities and once a certain temperature is achieved the billet is wetted with the metal melt poured into the mould gap (the metal melt is forming a cladding layer, which is a necessary condition for the welding-brazing process).

Processing of high-alloyed steel wastes, in particular, high-speed steel wastes, is one of the most effective trends in the electroslag technology applications. An important peculiarity of the electroslag technologies based on the use of the liquid metal is the use of not only spent machine parts or components of the metallic structures including different objects of military engineering [40, 41] for the melted material but also non-compact materials of relatively small fraction composition (chips, chops, granules, powder and others), which allows practically excluding the losses of tungsten, molybdenum, chromium, nickel and other scarce metals. In this case, as it has been indicated, there is no need in producing of consumable electrodes, which makes the process much cheaper and does not require complex technological equipment for implementation of the process. A new electroslag technology on the basis of the wastes utilization can be successfully applied in production of the alloyed billets including high-percent (65–75 %) ferrotitanium for which it is important [42] to provide a homogeneity of the ingots with minimal segregation and a low content of carbon and nitrogen.

The use of the EST LM opens fundamentally new possibilities for repair of rolls and continuous casting machine rollers. A doubtless advantage of the electroslag technology with a liquid metal lies in a possibility to surface the billets with the materials of various chemical composition, including high-chromium cast iron, high-speed steel and others.

In solving the strategic and tactic problems of the modification (reconstruction) of the electroslag equipment it is important in the first place to identify the basic range of finished and half-finished products to be commercially manufactured by the introduction of new electroslag technologies.

This range of products may include billets of high-speed and tool steels including the wastes utilization; bimetallic billets for production of plates, pipes and special reinforcement with corrosion-resistant coating; cast round billets including hollow billets of austenite steels and alloys for nuclear engineering,

oil chemistry and others; rolls and continuous casting machine rollers; hollow billets of steels and alloys for subsequent hot rolling on the ring-rolling mill; ingots of high-alloy steels and alloys containing high-reaction alloying elements (titanium, aluminium and others); alloying composition billets on the basis of the wastes utilization.

It should be, however, recognized that many ESR furnaces, especially one-phase, developed in the mid of 60s early 70s and supplied then to the plants and shops of the CIS countries, do not correspond to the modern level of furnaces for melting of ingots and billets of the mentioned range as well as to the requirements necessary for industrial implementation of the newly developed technological ESR processes. Though these furnaces used to be the best in the world as to their design, now they require modification (or reconstruction). In some cases they should be replaced with new modern furnaces.

In this relation the Novo-Kramatorsk Machine-building Plant (NKMZ Company), the leader of the Ukrainian machine-building and the world recognized producer of the rolling equipment may serve as a good example for domestic (and not only domestic) producers of the ESR metal. Three-phase ESR furnace for industrial production of forging ingots was for the first time put into operation at this plant: EShP-2 for producing 2.5 t ingots in 1959 and EShP-10 for producing 10–12 t ingots with simultaneous remelting of three consumable electrodes in one mould in 1962 [43]. These events pioneered the domestic and world production of large ESR ingots for manufacturing of turbine rotor necks, rolls for cold rolling and other critical components. In the next years in line with the new tasks and gained experience the ESR equipment was under continuous process of elaboration and renovation, which resulted in creation of the ESR shop largest in the heavy machine-building and unrivalled both in the former USSR and abroad.

Even today under economical constraints NKMZ Company has found the necessary facilities to carry out the project for adaptation of the 20-t ESR furnace and to create in the specialized shop an industrial unit for electroslag liquid metal surfacing [44–47]. The unit is designed for production of the composite roll billets up to 1000 mm in diameter and with the length of the surfaced body up to 2500 mm and up to 20 t mass by the method of ESS LM. A technology for surfacing of rolls for hot strip mill of 740 mm in diameter with a working layer of high-speed steel was adopted at this unit and the first lots of these rolls were produced. Trials of the composite ESS LM rolls on the wide-strip mill have exposed their 4–4.5 times increased resistance as compared to that of the previously used cast iron rolls with a chill layer. In the nearest future the composite ESS LM rolls will be ready to a wide entrance to the world market and become a serious alternative to the available composite rolls, in particular produced by the method of centrifugal casting [45].



Now NKMZ Company has put into operation the second more powerful furnace for production of the bimetallic rolls and other components up to 1600–1800 mm in diameter by the ESS LM method.

Therefore, modification (reconstruction) of the available electroslag furnaces and introduction of new electroslag technologies is the main condition for effective production of the ESR metal at the plants in Ukraine. This is the only way to increase the production volumes of the ESR metal, decrease of the costs and manufacturing of high-quality product compatible at the world markets.

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APPLICATION OF ELECTROSLAG TECHNOLOGY FOR REFINING TITANIUM AND TITANIUM ALLOYS FROM NITROGEN-RICH INCLUSIONS

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The feasibility of use of a chamber electroslag remelting for refining titanium alloys from nitrogen-rich inclusions was checked experimentally. It was established that the application of CaF_2 -Ca flux system makes it possible to create the necessary thermodynamic conditions for decomposition of nitrogen-rich inclusions during the process of a chamber electroslag melting.

Key words: *electroslag remelting, calcium flux, titanium alloy, nitrogen-rich inclusions, refining*

It is known that quality and service characteristics of titanium and titanium-base components with their increasing application areas depend on the purity of the material and on the availability and composition of non-metallic inclusions. The so-called nitrogen-rich inclusions (NRI) prove to be a serious hazard for physical properties of this material. These inclusions have, as a rule, a core of TiN_x surrounded with layers containing α -Ti and β -Ti. NRI known in literature as «hard» α -inclusions are brittle in their nature and under various loads may cause crack initiation in metal. Problem in this case consists not in the presence of nitrogen in the system, but in the susceptibility of NRI to cracking. Hard or brittle inclusions and pores, their combinations or other homogeneity disturbances are the centers of crack nucleation. The greater is the crack nucleation center, the larger is the initial crack — the greater is the fatigue crack growth velocity and the shorter is the time before failure under low-endurance fatigue [1, 2].

NRI are formed under production of titanium sponge and during its process stages and are not fully removed under melting [3, 4]. Grinding of spongy titanium into smaller parts [5] and application of melting techniques expanding the time period of its liquid stay (triple vacuum-arc, electron beam and plasma remelting) improve the situation but do not guarantee a complete removal of the inclusions. The problem, therefore, lies in removal or minimization of the size of NRI in titanium and titanium-base alloys.

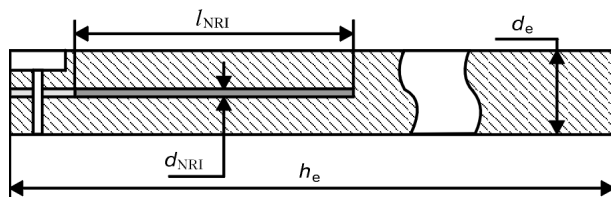


Figure 1. Diagram of the electrode with NRI in its axis

Analysis of the Ti-N constitution diagram shows that all NRI melt under higher temperature than titanium [6]. Besides, a density of titanium nitride TiN (5.43 g/cm^3) exceeds density of titanium (4.5 g/cm^3) and its alloys [7]. Hence, removal of NRI from the liquid titanium under electroslag remelting (ESR) by melting and floating seems improbable.

Study of dissolution velocity of NRI submerged into liquid titanium has shown that this process is very slow (about $2.2 \mu\text{m/s}$) [8], thus restricting decrease of the NRI size during their dissolution in the liquid titanium under canonic ESR.

Application of electroslag remelting under Ca-containing fluxes may prove thermodynamically favorable for removal of such inclusions [9]. This process provides a low (about 10^{-11} kPa) partial pressure of nitrogen in slag making it possible to refine titanium and titanium alloys from NRI [10].

This work reports the results of the experiments on removal of NRI under ESR of commercial titanium alloy VT6-4. This alloy is widely used as structural material in aviation, chemical and power machine-building. Usually it is produced from titanium sponge by the method of vacuum-arc remelting and subjected to full ultrasonic investigation. Metal with the detected NRI defects is rejected and sent back to a repeated process stage, which brings about large economical losses.

Electroslag remelting is conducted in the chamber electroslag furnace created on the basis of the unit A-550 [11], in the copper water-cooled mould 135 mm in diameter 500-mm long. Water-cooled chamber is installed directly on the upper part of the mould. A system is designed with corresponding sealant, which allows creating vacuum or excessive pressure in the working space. Before melting the chamber is evacuated and then filled with argon. An excessive pressure of argon (about 15 kPa) is maintained in the system to compensate losses caused by various seals.



Content (wt.%) of nitrogen and oxygen in the alloy VT6-4

Melt No.	[O]	[N]
Initial	0.17–0.18	0.010–0.020
1	0.17–0.22	0.020–0.029
2	0.21–0.22	0.026–0.030
3	0.18–0.19	0.033–0.041

Consumable electrodes ($d_e = 77$ mm) made of titanium alloy VT6-4 (Figure 1) with artificially prepared NRI introduced into their upper part are used for the experiments. NRI are made of TiN powder containing 49 at.% (22 wt.%) of nitrogen by pressing into the titanium alloy tube, which after sealing and EBW of face ends is subjected to isostatic pressing at the temperature 1475 K and pressure $2 \cdot 10^5$ kPa during 4 h. In the axes of the electrodes a hole was drilled to where the artificially prepared NRI ($d_{\text{NRI}} = 1$ mm) are placed. The hole is sealed with the titanium alloy plug and welded by argon-arc welding. The ready electrode was subjected to hot isostatic pressing (Figure 2). A construction of the artificial axial inclusion used in the electrode practically excludes its drop-out under remelting.

A powder of calcium fluoride of «Ch» grade (TU 6-09-5335–88) calcinated at the temperature 973 K during 3 h and metallic calcium are used as flux-forming materials. Flux is melted directly in the mould using the «hard» start. Calcium vapors formed under melting of CaF_2 –Ca decrease the partial pressure of nitrogen and oxygen over the slag pool down to the values 10^{-13} kPa.

Test melts are performed with similar electric parameters ($U = 47$ V, $I = 3$ kA) providing a satisfactory quality of the surface formation in the ingots equal in mass. A flux composition (CaF_2 or $\text{CaF}_2 + \text{Ca}$) and a length of the melted part of the electrode with the introduced artificial NRI are changed.

A melt No.1 (see the Table) under pure calcium fluoride (2300 g mass) is chosen as a standard one. Length of the fused part of the electrode with NRI is $l_{\text{NRI}} = 125$ mm. Melts No.2 and 3 differ from the standard one by addition of metallic calcium (100 g) to the calcium fluoride and by the length of the fused parts of the electrodes with NRI — 35 and 85 mm, respectively.

Test ingots are subjected to ultrasonic investigation by a Krautkramer Branson device USN 52 at the frequency 5 MHz, layer facing on the turning machine and visual control. Gas content in the metal is detected at the LECO device in the carrier gas flow in



Figure 2. Appearance of the electrode of titanium alloy VT6-4 with NRI

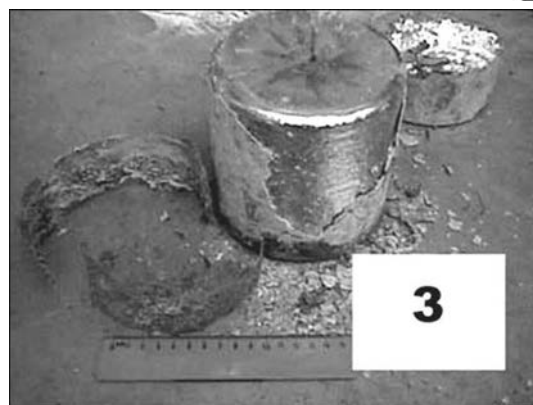


Figure 3. ESR ingot of titanium alloy VT6-4 (melt No.3)

the specimens from the transversal templates 200 mm thick cut from the head of the ingots.

Examination of the surface of the produced ingots testifies to the fact that additions of the metallic calcium to the calcium fluoride sharply decrease a residual content of oxygen and nitrogen in the melting zone. This is proved by a vividly expressed silvery (non-oxidized) color of the surface of ingots (melts No.2 and 3) as compared to the dark goldish color of melt No.1 ingot. Moreover, under stripping of ingots No.2 and 3 self-ignition of the slag fumes on the sur-

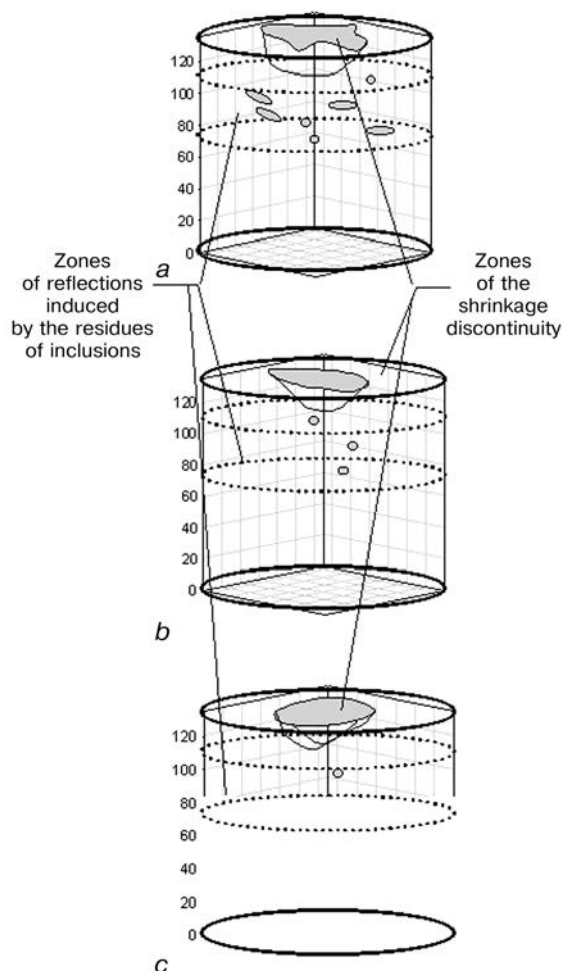


Figure 4. Diagram of location of the ultrasonic wave reflections: a — in the melt No.1 ingot; b — in the melt No.3 ingot; c — in the melt No.2 ingot

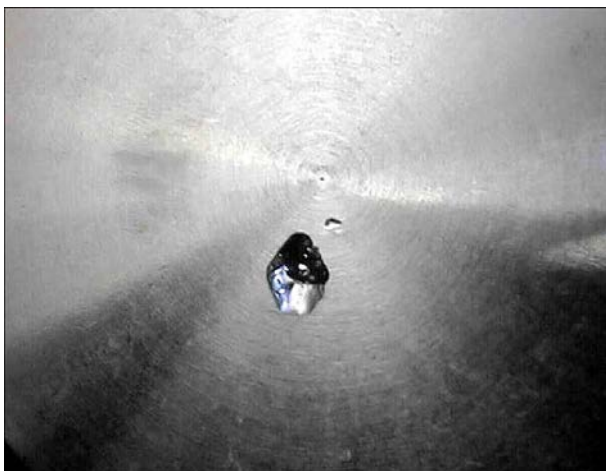


Figure 5. Shrinkage pores



Figure 6. Residues of the undestroyed NRI (ingot No.1)

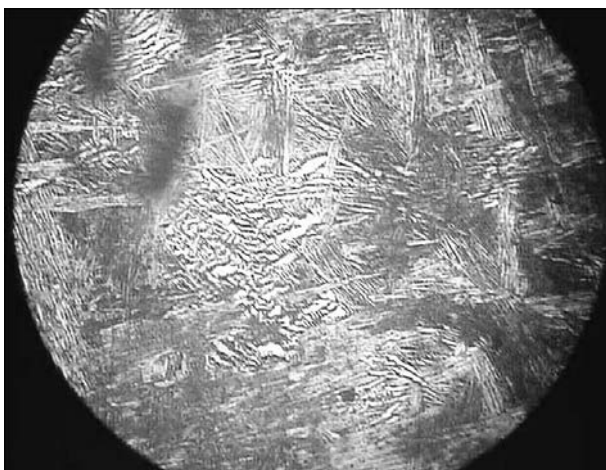


Figure 7. Nitrided zones in ingot No.3



Figure 8. End face of the electrode with protruding NRI (melt No.1)

face of the mold is observed, which is caused by a surplus of the metallic calcium vapors in the melting zone. This indirectly proves the fact that additions of metallic calcium to calcium fluoride create a deeply deoxidized medium.

Ultrasonic investigation of the test metal detects the presence of two zones of localized reflections in all ingots (Figure 4). The first zone is located in the upper part of ingots and is characterized by multiple reflections concentrated in their central axis. An area occupied by these reflections and their location in different ingots are of similar nature. Extension of these zones by the height of the ingot does not exceed 20–30 mm from the upper face. Based on the amplitude and localization of the reflections it is assumed that they are induced by the shrinkage defects. The second localization zone of the defects is located at a distance of 35–65 mm from the upper faces of the ingots. The number of reflections in them is different for different melts. The maximal number of signals is recorded in the standard ingot. In ingot No.2 there are almost no reflections in this zone. The number and amplitude of the reflections in ingot No.3 is essentially lower than in the standard.

The results of ultrasonic investigation are checked by machining of ingots in the turning machine: step by step metal layers are removed by facing and the



Figure 9. End face of the electrode (melt No.2)



exposed surface is viewed. It is established that upper signs of ultrasonic investigation (Figure 4) correspond to the shrinkage pores with clean non-oxidized inner surface (Figure 5). Multiple residues of the undestroyed NRI are observed in the middle part of ingot No.1 (Figure 6). Echo signals in the ingots No.2 and 3 correspond to the nitrated zones (Figure 7) detected metallographically. They are the sites of solid solution of nitrogen in titanium and are sharply different in their composition and structure from the artificially introduced equiatomic titanium nitride. Negative impact of these sites on the mechanical characteristics of the alloy (particularly on fatigue strength) is considerably lower than the impact of TiN particles, which sharply differ in structure, physical and mechanical properties from the basic alloy.

Gas content in metal, particularly the content of nitrogen in the alloy VT6-4 is determined with the aim of control of possible processes of decomposition of NRS under remelting using different slag. As it is seen from the Table, the content of nitrogen in the metal of ingot No.1 increases insignificantly in despite of the larger quantity of the introduced NRI (125 mm) as compared to ingots No.2 and 3. Additives of metallic calcium into the slag under chamber ESR result in the increase of nitrogen content in the metal, i.e. the larger is the amount of the added metallic calcium, the greater is the mass of the introduced NRI (melts No.2 and 3). This testifies to the fact that during chamber remelting under Ca-containing slag the necessary thermodynamic conditions for NRI decomposition and transfer of nitrogen into the ingot metal are created. The oxygen content in the titanium alloy is practically unchanged under considered remelting variants.

A nature of the fusion of electrodes with the introduced axial inclusions confirms the conclusion on favorable thermodynamic conditions of the NRI decomposition under application of the Ca-containing fluxes.

So, under melt No.1 the electrode melting rate considerably exceeds the decomposition rate of the NRI rod, which by these means reaches the slag-metal interface, induces short-circuits and destroys mechanically precipitating in the metal pool (Figure 8). Fusion nature of electrodes under melts No.2 and 3 is different (Figure 9). It is seen that the rates of electrode melting and NRI decomposition are comparable in these cases.

Therefore, the results of the studies have proved appropriateness to apply the chamber electroslag remelting under Ca-containing flux for refining titanium and titanium-base alloys from the NRI. Implementation of this technology will permit returning to the production cycle considerable quantities of previously rejected titanium alloys.

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EFFECT OF ELECTRON BEAM REMELTING ON STRUCTURE AND PROPERTIES OF HIGH-SPEED STEEL R6M5

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Results of examination of physical-mechanical properties, microstructure, chemical and phase compositions of high-speed steel R6M5, produced by electron beam remelting of industrial waste, are given. The decrease in chemical inhomogeneity, high dispersity and uniformity of distributed carbide phase in the produced EBR ingots are outlined. The life of a tool manufactured from the steel examined 1.1–1.4 times exceeds the life of a tool from high-speed steel (forgings) of a traditional melting.

Key words: *electron beam remelting, high-speed steel, ingot, microstructure, defects, grain, carbides, carbide network, mechanical properties, heat treatment, red-hardness*

This work is devoted to the study of macro- and microstructure, chemical and phase composition as well as certain physical-mechanical and service properties of high-speed steel R6M5 produced from industrial wastes of tool production by the electron beam remelting (EBR) technology developed at the Research-Production Enterprise «GEKONT» (Ukraine) [1, 2] and tools made of this steel. The results have shown that the «crude» ingots of high-speed steel R6M5 of EBR (Figure 1) are distinguished by high hardness, especially on the surface because steel undergoes martensite transformation (hardening) under high rate of solidification and cooling of metal in the water-cooled copper mould. Therefore, for manufacturing of tools the produced ingots are subjected to high-temperature spheroidizing annealing by the scheme: slow heating up to 850–870 °C with 3-hour holding, cooling down to 730–750 °C with 6-hour holding and then slow cooling with furnace.

Study of the EBR ingots macrostructure on the longitudinal templates has shown that the macrostructure is dense without porosity, undermelting, flaws, discontinuities, cracks and shrinkage defects.

Chemical composition and distribution of alloying elements in the section of ingots is homogeneous and meets the requirements of GOST 19265–73 (Table 1). Inconsiderable relative increase of the carbon, tungsten, molybdenum and vanadium content in the ingots as compared to the wastes for remelting is attributed

to the loss of metal, in particular iron, chromium, manganese and copper because of their evaporation in vacuum under the action of electron beams. A decrease in the content of sulfur, nitrogen and oxygen in steel is also observed. Total losses of metal due to its evaporation are 0.5 wt.% under remelting.

Longitudinal specimens for mechanical tests and transversal specimens from top, middle and bottom parts of the ingots for structural and durometric analysis are cut from «crude» and annealed ingots. For hardening the specimens are heated in the salt pool containing 95 % BaCl₂ and 5 % MgF₂ within 1190–1220 °C with cooling in the alkaline pool at the temperature 450–500 °C. The hardened specimens are subjected to two-fold salt tempering for 1 h at the temperature 560 °C.

Tensile strength tests are performed at the room temperature on the universal testing machine 1246-R with the displacement rate of the active machine grip 2 mm/min using ten-fold cylindrical specimens with a working part diameter 5 mm (GOST 1497–84).

Specimens after hardening with subsequent tempering are exposed to brittle failure (without plastic deformation). Visual observations show that this failure is premature (the material has not exhausted its ultimate strength) and is attributed both to high brittleness of the material and to difficulties of the tests (it is impossible to completely get rid of eccentric application of forces). Impact bending tests (impact strength) are performed on the specimens with a V-shape concentrator in line with GOST 9554–78. The results of mechanical tests are presented in Table 2.

Relatively low impact strength of specimens is explained by the presence of carbide network, which can be removed only by applying a complex and labor-consuming thermomechanical treatment (forging + annealing). Only annealing at increased temperature and holding without forging results in a considerable enlargement of austenite grain and development of carbide heterogeneity [3].

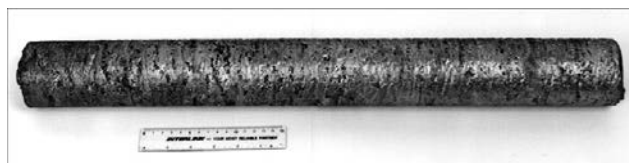


Figure 1. Appearance of the ingot 70 mm in diameter from high-speed steel R6M5 produced by EBR

**Table 1.** Chemical composition of high-speed steel R6M5 before and after EBR

Object of investigation	Content, wt. %											
	C	Mn	Si	Cr	W	V	Mo	Cu	S	P	[O]	[N]
R6M5 (GOST 19265-73)	0.82– 0.90	0.20– 0.50	0.20– 0.50	3.80– 4.40	5.50– 6.50	1.70– 2.10	4.80– 5.30	≥0.25	≥0.025	≥0.030	–	–
Scrap (average chemical composition)	0.85	0.34	0.31	4.30	5.90	1.80	4.86	0.24	0.018	0.021	0.023	0.009
Ingot (bottom part)	0.89	0.31	0.30	3.80	6.10	1.86	4.90	0.17	0.009	0.017	0.015	0.006
Ingot (top part)	0.91	0.28	0.28	3.84	6.20	2.00	5.10	0.19	0.011	0.019	0.018	0.004

Table 2. Mechanical properties of high-speed steel R6M5 of EBR

Specimens	$\sigma_{0.2}$, MPa	σ_l , MPa	ϕ , %	KCV, J/cm ²	HRC	Character of failure
Cast	696	816	1.6	8–10	52–54	–
After annealing	768	771	0.8	12–18	24	–
After heat treatment (hardening + tempering)	–	691	–	4–8	63–64	Brittle
After annealing and heat treatment	–	776	–	6–10	65–66	Same

Fractographic studies of fractures in specimens destroyed under mechanical tests as well as microstructure investigation are performed on the raster electron microscope REM-200. A phase composition is studied by the results of X-ray diffraction analysis of crystal structure of specimens in the diffractometer DRON-4 in the characteristic cobalt radiation.

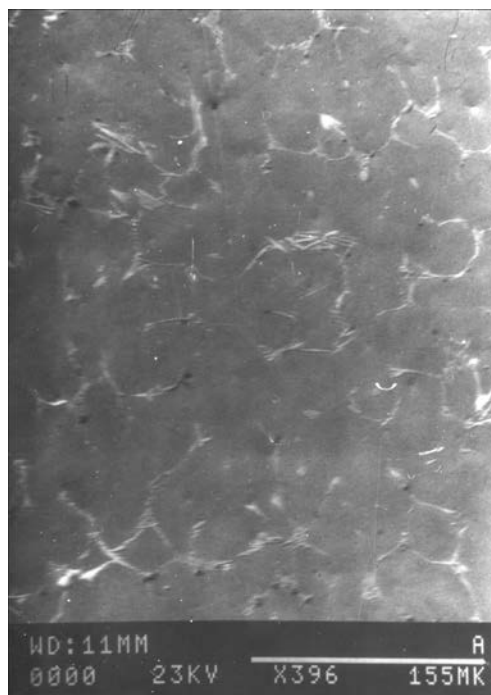
Analysis of microstructure and a nature of failure of certain sites in the specimens allows detecting some peculiarities of structure and properties of high-speed steel R6M5 of EBR. Microstructure of the transversal section produced from the «crude» ingot consists of globular martensite grains (grain index is 8–9), residual austenite, broken carbide network (6–7 index of carbide heterogeneity according to GOST 19265–73, scale 2) on the boundaries of grain and disperse carbides uniformly distributed inside the martensite grains (Figure 2).

Carbide size is 3–5 μm , thickness of carbide network is 2–7 μm . The number of residual austenite is determined by means of a magnet austenometer MAK-2M in the calibrated standard of steel R6M5. Its fraction for cast steel is 16–19 vol.%. The electron microfractogram of the destroyed specimen of cast steel (Figure 3, *a*) shows a mixed brittle-viscous nature of failure (transcrystal and intercrystal shear + «cup» fracture). Aside from eutectic carbides M_6C having a dendrite morphology and forming a carbide network around austenite grains (Figure 2) the X-ray analysis of specimens shows 5–10 vol.% of metastable carbides M_2C (globular shape) and inconsiderable quantity of primary vanadium carbides (VC) as irregular polyhedrons (see Figure 3, *a*) in the structure of the cast steel. This can be attributed to an increased concentration of vanadium because of selective evaporation of elements and small overheating of the melt as well as to a high rate of solidification and cooling of the melt. One may also assume that vanadium

carbides are the centers of nucleation under melt solidification.

Analysis of the results obtained under study of specimens after annealing and hardening with subsequent two-fold tempering allows noting the following peculiarities of the steel R6M5 of EBR.

Structure of the annealed steel consists of globular sorbitic pearlite (grain index 8–9), residues of the carbide network on its boundaries and dispersed carbides. A total fraction of carbides decreases down to 10–12 vol.% as against cast steel because of their partial dissolution in austenite under holding at the temperature above 800 °C. This factor as well as a high concentration of alloying elements in the solid

**Figure 2.** Microstructure of cast high-speed steel R6M5 of EBR

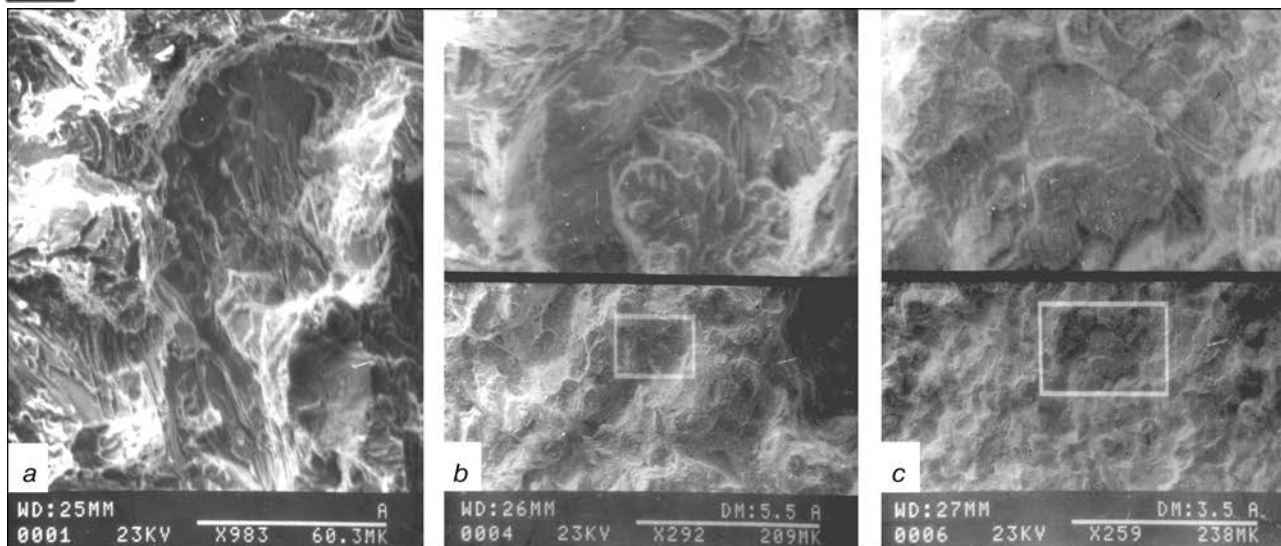


Figure 3. Electron microfractograms of fractures of high-speed steel P6M5 of EBR: *a* — cast metal; *b* — after annealing; *c* — after hardening and two-fold tempering

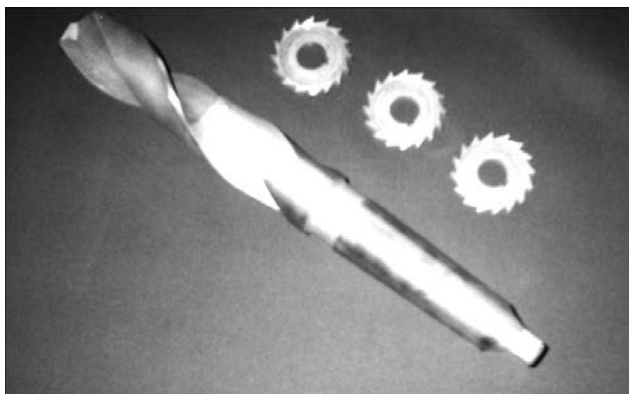


Figure 4. Tools of high-speed steel P6M5 produced by EBR of industrial wastes

solution account for a somewhat increased hardness of the annealed steel (HB 250–270) and its relatively low impact strength [4]. Failure also proceeds by a mixed mechanism with a decrease of a quantity of the facets of shear (increase of a fraction of a viscous component of the failure) (Figure 3, *b*).

Structure of high-speed steel R6M5 of EBR after hardening and two-fold annealing (Figure 3, *c*) consists of globular high-alloyed martensite (grain index 9–10), broken carbide network (index 6–7) and globular carbide: metastable M_2C and high-stable VC whose total volume fraction is about 15–18 % of carbide phase. The quantity of austenite is inconsiderable — up to 4 vol.%. Failure proceeds by a mixed mechanism of dispersed quasishar with inconsiderable fraction of viscous «cup» fracture. Steel hardness is high — HRC 64–66. Red-hardness of steel (tempering resistance) corresponds to HRC 59 at the temperature 620 °C for 4 h.

Annealed ingots of high-speed steel R6M5 of EBR are used for production of cutting parts of drills 60 mm in diameter without shanks and side mills 60–80 mm in diameter (Figure 4) at the Company «Vinitsky Instrumentalny Zavod». After standard heat treatment (hardening and three-fold tempering) tools were tested in the following conditions:

- drills 60 mm in diameter — feed rate is 0.56 mm/rev, cutting rate is 40 m/min, the number of revolutions is 140 rpm under drilling of carbon steel 45 GOST 1050–88 (hardness HRC 190), average resistance is 72 min;
- side mills 63 and 80 mm in diameter — cutting rate is 40 m/min, feed rate is 0.06 and 0.08 mm/tooth, the number of revolutions is 200 and 160 rpm for the same steel, average resistance of tool is 68 and 74 min, respectively.

Results of the industrial tests have shown that resistance of tools from high-speed steel R6M5 of EBR of industrial wastes is 1.1–1.4 times higher than the resistance of tools produced from high-speed steel (forging) of traditional melting.

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AMPLITUDE DEPENDENCE AND MICROSCOPIC ELASTICITY LIMIT OF THE Ni-63 % Co MATERIAL DEPOSITED FROM THE VAPOR PHASE

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The amplitude dependence of internal friction was investigated, and the microscopic elasticity limit of composite condensates Ni-63 % Co produced by electron beam evaporation and subsequent condensation in vacuum was determined. The role of the mechanism of microplastic deformation in formation of high damping ability of the material under investigation was evaluated.

Key words: electron beam evaporation, damping factor, microscopic elasticity limit

Experimental study of initial stages of plastic deformation has shown that the latter is conditioned by the processes proceeding under microplastic deformation. A specific substructure of metals and alloys determining behavior of the material under subsequent loading is formed at the stage of microplasticity.

It is known that Ni-Co alloys possess a high level of damping. In Cochardt's works [1, 2] it is noted that damping in ferromagnetic materials is mainly conditioned by the magneto-mechanical hysteresis, i.e. irreversible shift of the domain boundaries under alternating load.

Earlier the amplitude dependence of Ni-Co alloys was studied in [3]. In particular, an amplitude dependence of the alloy containing 63 % Co was investigated. A non-monotonous dependence of the inner friction on the deformation amplitude was established. A peak was observed on the curve of the amplitude dependence at the room temperature in the absence of magnetic field under deformation amplitude $\gamma = 4 \cdot 10^{-4}$. The authors attribute the presence of peak to two processes — irreversible shift of the domain boundaries at $\gamma \approx 10^{-5} - 4 \cdot 10^{-4}$ and microplastic deformation occurring at $\gamma > 4 \cdot 10^{-4}$. In [3] it is also shown that under deformation amplitude $\gamma = 2 \cdot 10^{-5}$ damping only slightly depends on the value of the constant magnetic field. The above suggests that under lower deformation amplitudes the damping does not depend on the value of magnetic field.

In order to identify a role of microplastic deformation in damping an amplitude dependence of the internal friction under small deformation amplitudes $1.3 \cdot 10^{-6} - 1.7 \cdot 10^{-5}$ is studied and a microscopic elasticity limit τ_s of nickel condensate containing 63 % Co is determined.

Composite material Ni-63 % Co is produced by electron beam evaporation with subsequent condensation in vacuum [4]. Specimens $\sim 0.3 \times 3$ mm in section with a working part 24 mm long are made of this material. Before tests the specimens are annealed in vacuum at the temperature 850 °C for 1 h.

A logarithmic damping factor δ_0 of torsional freely damped oscillations is assumed as a value of internal

friction. Measurements are performed at room temperature in the air. A frequency of torsional oscillations is ~ 0.6 osc/s.

An amplitude dependence of the damping factor obtained under successive increase of the initial deformation amplitude from $1.3 \cdot 10^{-6}$ to $1.7 \cdot 10^{-5}$ is presented in Figure 1. As it is seen, the damping factor increases under rather small amplitudes $1.3 - 1.4 \cdot 10^{-6}$. Within the range of amplitudes $10^{-6} - 10^{-5}$ this dependence has a linear character. Starting from the amplitude $\gamma = 10^{-5}$ the inner friction sharply increases. A one-fold increase of the deformation amplitude results in almost 5-times increase of δ_0 .

Microscopic elasticity limit is determined as a minimal shear stress under which the value of the damping factor increases, i.e. irreversible plastic losses in the microdomains take place. Inner friction is measured simultaneously under torsion of the specimen within its macroelasticity by the method described in [5]. An increase rate of the additional tangential stress is 23 g/(mm²·s).

Within the indicated ranges of the measured amplitudes the damping factor depends both on the deformation (stress) amplitude of torsional oscillations and on the applied additional shear stress. Therefore, two competing processes adversely affecting the energy dissipation simultaneously proceed in the material.

First, inner friction decreases together with the deformation amplitude. Second, when additional

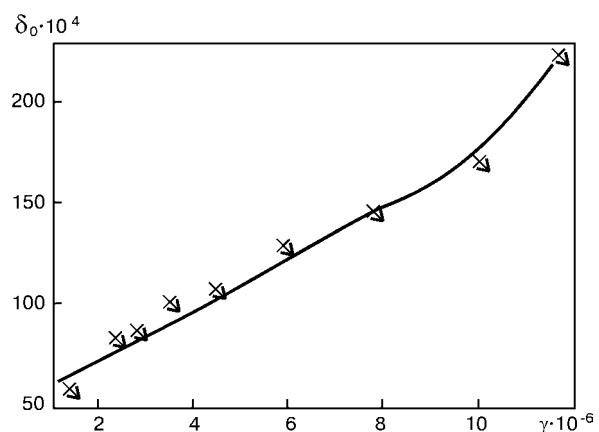


Figure 1. Dependence of the logarithmic damping factor of condensate Ni-63 % Co on the deformation amplitude

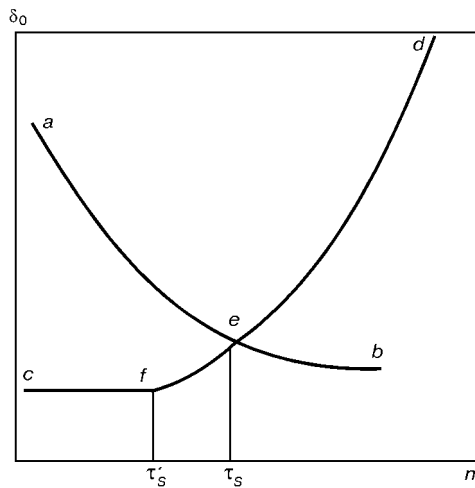


Figure 2. Curves of the dependence of the damping factor on the number of oscillation: curve *ab* — under conditions of the amplitude dependence in the absence of additional stress; curve *cd* — in presence of additional stress in the amplitude-independent domain

stress achieves a value of microscopic elasticity limit the inner friction grows. This leads to a complex dependence of the damping factor of the stress applied to the specimen.

Hence, under application of the alternating stress the inner friction decreases with the growth of the number of oscillations while under simultaneous application of both alternating and additional shear stress the inner friction increases. This is illustrated in Figure 2 where values of the damping factor are laid off along the axis of ordinates and the number of oscillation n — along the axis of abscissae. The curve *ab* corresponds to the dependence of the damping factor on the stress amplitude of the torsional oscillations, the curve *cd* — to the change of the damping factor under application of additional stress in the absence of the amplitude dependence. Stress in the point *f* is equal to a microscopic elasticity limit. Both processes proceed simultaneously until local stresses resulted from the irreversible microplastic deformation are sufficient to block the shift of the boundaries of domains. As it is seen from the Figure 2, the curves cross in the point *e*. The obtained curve *aed* shows a character of the damping factor change in the process of monotonous increase of additional tangential stress under conditions of the amplitude dependence. A descending branch of the curve is determined mainly by the amplitude dependence, an ascending one is conditioned by irreversible microplastic losses. A minimum position depends on the correlation of the development rates of these processes. The point *e* corresponds to the microscopic elasticity limit.

For comparison the results of the measurements of the inner friction under increase of the additional stress in presence of the amplitude dependence are shown in Figure 3. As it is seen from the graph, the microscopic elasticity limit proves to be 385 g/mm^2 , i.e. it equals approximately $1/12$ of the static elasticity limit (the static elasticity limit is determined under tension and in terms of tangential stress it equals 4.84 kg/mm^2). The value of the microscopic elasticity limit allows evaluating the length l of the

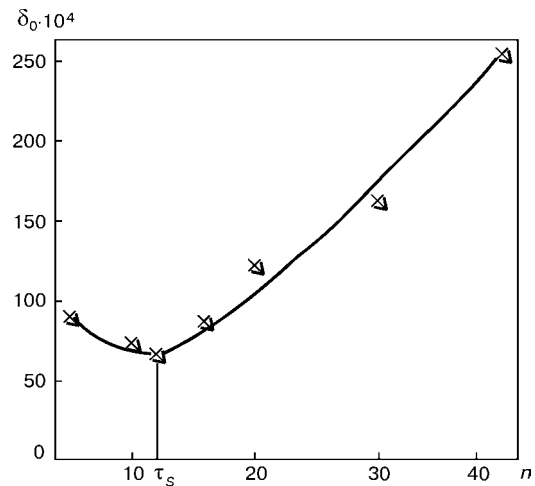


Figure 3. Dependence of the damping factor of condensate Ni-63 % Co on the number of oscillations in presence of monotonously increased stress under conditions of the amplitude dependence

sources capable to generated dislocation in compliance with the dependence

$$l = \frac{GB}{\tau_s},$$

where G is the shear modulus; B is the Burgers vector.

The obtained value $l = 4.8 \text{ } \mu\text{m}$ coincides with the size of the condensate grain ($\sim 5 \text{ } \mu\text{m}$) after annealing at 850°C for 1 h. Additional loading under described conditions leads to hardening in consequence of which τ_s achieves the value 735 m/mm^2 , i.e. increases almost twice.

The value of τ_s obtained experimentally may not correspond to the calculated value of τ'_s , i.e. if the energy dissipation conditioned by the amplitude dependence proves higher than the dissipated energy related to the losses as a result of irreversible plastic deformation in microdomains within the range limited by the values τ_s and τ'_s . Then the measured value τ_s will be somewhat higher than τ'_s .

However, this difference cannot be considerable since even after 200 oscillations the amplitude dependence still takes place while an increase of the inner friction under joint action of periodic and increasing shear stress appears already after several oscillations.

CONCLUSIONS

1. A value of τ_s under conditions of amplitude dependence of freely damped oscillations for the condensate Ni-63 % Co is determined.
2. The value l calculated by the value τ_s satisfactorily correlates with the size of the grain.

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HEAT MODEL OF PLASMA-ARC PROCESS OF GROWING TUNGSTEN SINGLE CRYSTALS*

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Heat model of plasma-arc process of growing rods of tungsten single crystals is presented. Results of numerical solution of heat conductivity equation are given. It is shown that with increase in single crystal height the power is increased while the pool parameters are not almost changed. It was established that the heat power is stabilized at the single crystal height equal to its three diameters.

Key words: tungsten single crystal, plasma-arc zonal melting, heat model

It is most rational to produce the oriented single crystals of refractory metals by using different methods of a zonal melting. One of them is the plasma-arc zonal melting (PAZM).

A plasmatron of direct or indirect action serves as a heat source in plasma-arc process of growing single crystals. Direct-action plasmatrons are used in a technological equipment for growing single crystals of the refractory metals. The plasmatron arc column blown with a flow of the plasma gas (argon and helium mixture) forms a plasma flame which looks like a bell at a visual observation. In the plasma arc transverse section regions of a different intensity of radiation and, as a consequence, of different temperature can be observed. The region adjacent to the plasma flame axial part, in which an arc column is located, is mostly heated. The column cross section area is changed depending on the cooling conditions which are controlled by the gas composition and degree of turbulization of the flow. The main fraction of current is passing here. From the literature data [1, 2] it amounts to about 98 % for the open arc.

Arc column is supported on electrode spots. The area of each spot is less than the area of the arc column cross section. Taking into account that the distribution of temperatures in arc cross section is close to that of Gauss [3], then it can be assumed that the energy density in a spot is not constant. Therefore, on the curve of distribution of temperatures for refractory metals a convexity in the arc spot middle is possible. The curve apex can be not so acute as the temperature of tungsten melting exceeds approximately 2 times the temperature of melting metals used earlier in the experiments [4]. From different estimations the temperature in the center of the plasma arc generated by a plasmatron is within 5000–16000 K [5–9]. Elasticity of vapors and overheating tempera-

ture of refractory metals are lower and, consequently, the degree of arc spot screening is lower.

If to adopt this as a basis then it can be considered that curve of distribution of temperatures on the pool surface heated by plasmatron can approach the Gauss distribution, but unlike the curve of distribution of temperatures in heating by electron beam, it has less acute apex and larger width at similar power. Taking into account the above-described, and also method of formalization of sizes of effective diameter of heating spot [4], accepted in electron beam heating, for modelling heat processes in growing single crystals with use of the plasma-arc source of heating it will be assumed that area of heating with plasma arc consists of two regions: central and periphery. The central area occupies the smaller part of the total area and has higher temperature. The limiting value of the temperature should not be higher than the metal boiling temperature. The periphery region has the temperature exceeding that of metal melting by several hundreds of degrees. In a perfect case the temperature of pool surface within each zone is similar in all the points.

The hypothesis confirmation was reflected in a real investigation which showed [4, 5] that there are two regions at the pool surface with a temperature differed by several hundreds of degrees. Moreover, the temperature transition between regions has not a smooth, but a jumpy nature. The curve of distribution of temperatures at the pool surface, and, consequently, the specific heat capacity, can be presented as follows:

$$\begin{cases} q = Q & r = 0, R_a; \\ q = hQ & r = R_a, R_l, \end{cases} \quad (1)$$

where q is the specific heat flow entering the metal in the arc heating zone; Q is the specific heat flow entered the metal in the arc spot zone; h is the normalization factor; r is the current value of radius in the arc heating zone; R_a radius of arc spot; R_l is the

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radius of zone of a metallic pool heated by the plasma arc.

In a real process the intensity of heat exchange between the plasma arc and metal depends also on the curvature of the metal pool surface which is formed as a result of effect of the arc gas-dynamic pressure on metal. There are attempts to determine the effect of surface curvature on the heat exchange processes [10–12]. However, this phenomenon is quite complex and represents a self-sustaining problem of the mathematical physics. In case of this phenomenon consideration, some difficulties are encountered in further obtaining and use of numerical or analytic solution. The accuracy of obtained results of heat modelling of the object will increase negligibly, while hardware difficulties will many times increase. Therefore, to analyze the heat processes proceeding in grow-

ing single crystals the effect of arc pressure on metal surface will not be accounted.

To construct the heat model of the process one formalization of the plasma-arc heat source is not sufficient. Distribution of temperatures in a crystal body will be determined adequately, if to preset initial and boundary conditions for non-stationary two-dimensional equation of heat conductivity:

$$\rho c_v \frac{\partial T}{\partial t} = \frac{1}{r} \frac{\partial}{\partial r} \left(r \lambda \frac{\partial T}{\partial r} \right) + \frac{\partial}{\partial z} \left(\lambda \frac{\partial T}{\partial z} \right), \quad (2)$$

where t is the current time; T is the current temperature; $\rho = \rho(T)$ is the metal density; $c_v = c_v(T)$ is the isochoric heat content of mass unit; $\lambda = \lambda(T)$ is the coefficient of metal heat conductivity. We shall consider a definite scheme of plasma-arc heating in growing single crystals of a cylindrical shape using the method developed at the A.A. Bajkov Institute of Metallurgy of Russian Academy of Sciences. The peculiar feature of the method is that the plasmatron and crystal are coaxial and the axis is arranged vertically. The possible scheme of modelling heat field of a single crystal is presented in Figure 1. The axis of a single crystal coincides with the axis of coordinates OZ . Due to a symmetry of the object a half of section of the single crystal is illustrated in the Figure. For the process modeling we shall assume the following conditions:

- heat flow from plasma-arc source enters the single crystal through its upper edge. This heat flow is resulting and accounts for all the heat processes: heat transfer from arc owing to convection and radiation; heat evolution in anode spot of arc; heat consumed for vapour formation; heat, evolving in condensation of metallic vapour at the pool surface; radiation of the metal pool into a furnace working space;
- distribution of density of heat flow is preset by equation (1). In the process of modelling the normalization factor can take any values within the interval (0.1);
- heat at the lateral surface of crystal and at its lower end is transferred by convection and radiation into the furnace space;
- at the initial moment of time the temperature of a single crystal is equal to the temperature of environment $T|_{t=0} = T_0$.

Boundary conditions will have the following form:

- at the crystal axis $-\lambda \frac{\partial T}{\partial r} \Big|_{r=0} = 0$;
- at the external surface of cylinder S , except the area S_a , $-\lambda \frac{\partial T}{\partial n} \Big|_S = Q_l + Q_c$, where $\bar{n}$ is the normal to surface; $Q_l = \sigma_B \left(\eta_S T^4 \Big|_S - T_0^4 \right)$ are losses in rays; η_S is the coefficient of heat radiation; σ_B is the Stefan-Boltzmann constant; $T \Big|_S$ is the temperature at surface S , except area S_a ; $Q_c = k_N \left(T \Big|_S - T_0 \right)$ are the

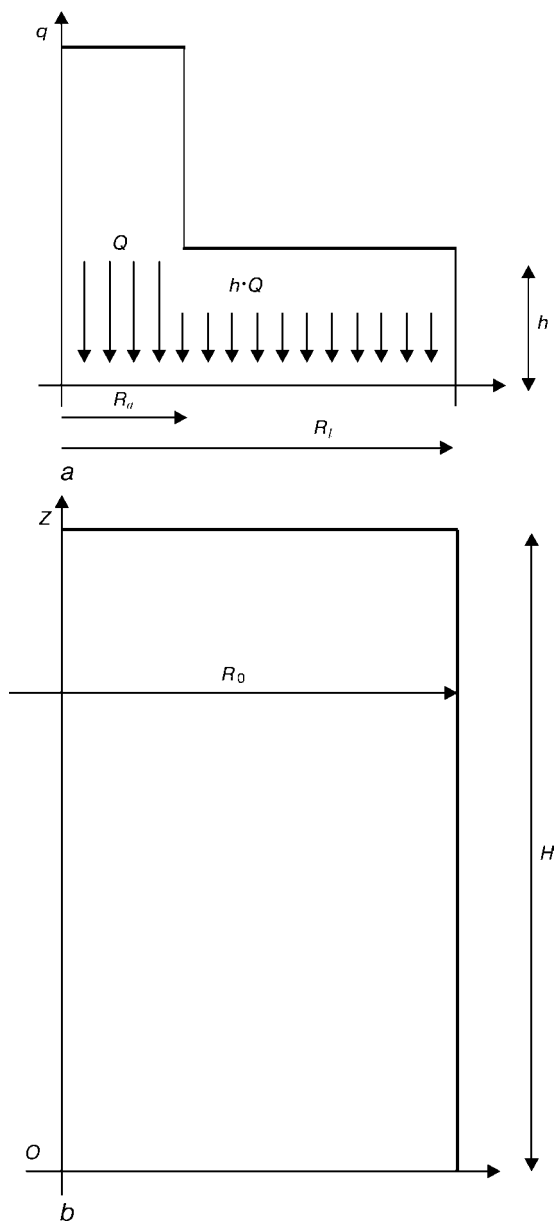
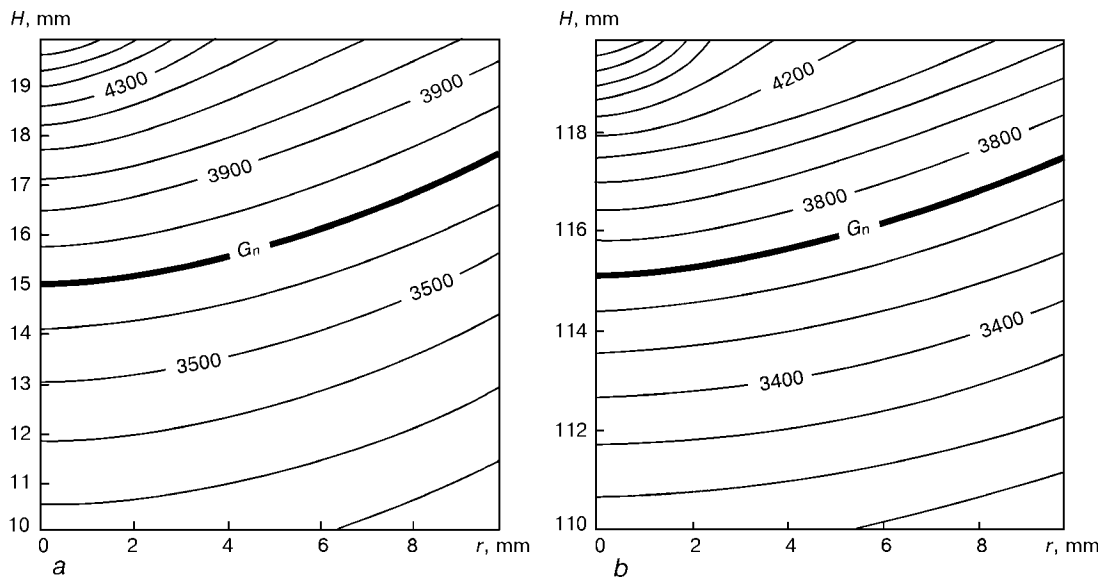


Figure 1. Scheme of modelling of heat field of a single crystal: a — heat flow entering the metal in the region of arc heating; b — single crystal in a cylindrical system of coordinates (H , R_0 — height and radius of single crystal, respectively)



convective losses; k_N is the coefficient of heat exchange with environment (Newton's factor);

- at the surface, on which the arc spot is supported in area S_a , $T|_{S_a} = T_b$, where T_b is the metal boiling temperature;

- at the surface of the phase transition G_n (solid/liquid phase — h/L)

$$-\lambda_L \text{grad} T \Big|_{G_{n(L)}} + \lambda_h \text{grad} T \Big|_{G_{n(h)}} = \rho \epsilon_L \vartheta_L,$$

where $\lambda_{L, h}$ is the heat conductivity of liquid and solid phases; ϵ_L is the energy of melting; ϑ_L is the rate of interface boundary movement.

Numerical solution of equation of heat conductivity is presented in the form of graphical relationships in Figure 2. The crystal growing using PAZM is featured by that the metal pool is maintained from flowing out by the surface tension forces. Therefore, the condition of constancy of melt height at the pool lateral surface and constancy of diameter of the crystal being grown were kept in modelling. It was determined experimentally that the height of melt at the lateral surface is usually within 2–2.5 mm. In modelling the melt height was taken equal to 2.5 mm. Modelling could reveal some laws of changing the power of plasma-arc heating in the process of the crystal growth. The power is changed from minimum value, at which the process is started on the priming crystal, almost to maximum at the crystal height equal approximately to three its diameters (Figure 3). Further, the results of modelling showed that it is necessary to increase the power, but this increase is so negligible in practice that it is very difficult to realize it from the technical point of view. Therefore, when the crystal reaches the height equal to three its diameters the power of heating is stabilized and remains unchanged until the process completion.

Interesting results were obtained on estimation of the pool depth (Figure 2, *a, b*). In spite of increase in heat power, the pool parameters are changed negligibly with increase in a single crystal height. This

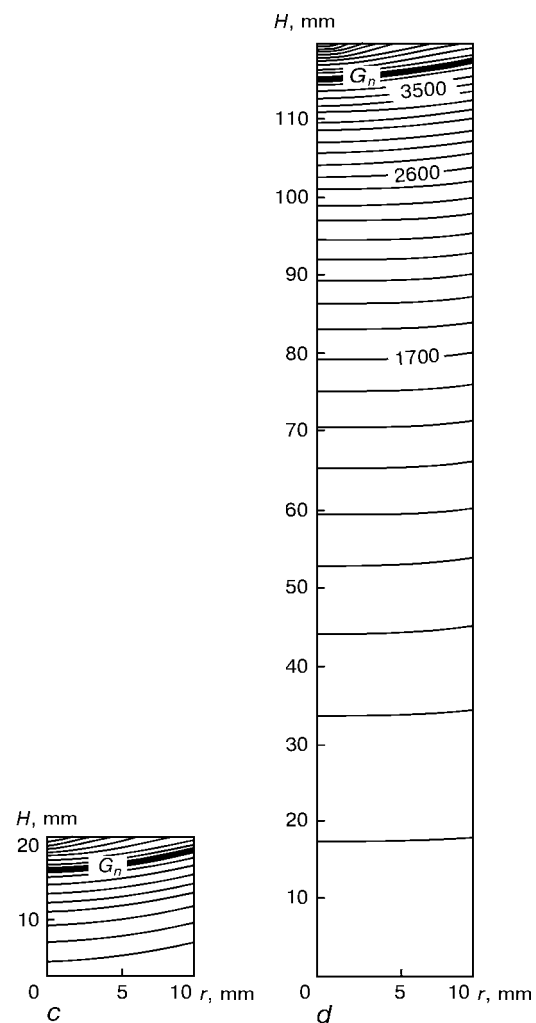


Figure 2. Temperature fields, K, in tungsten single crystals of height of 20 (*a, c*) and 120 (*b, d*) mm: *a, b* — region of metal pool; *c, d* — longitudinal section

makes it possible to produce the subgrain structure, not changing in the single crystal height. Curvature of the surface separating solid and liquid phase is also changed negligibly. The front of crystallization remains concave both at the beginning of the process of growing and at its end. The concave front of crys-

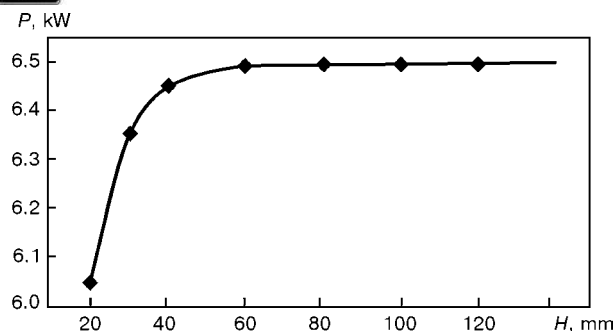


Figure 3. Change in capacity of plasma-arc heating depending on height of a tungsten single crystal (20 mm diameter rod)

tallization predicts that normals of local regions of solid/liquid phase interface will be oriented under the angle to the crystal axis and directed inside. They predetermine the direction of growth (final filling) of crystal lattice, but not the crystallographic direction which is not changed. The direction of subgrains is also not changed. With crystal growth the joining of several subboundaries is possible on the axis. Ternary butts are especially dangerous. They are weakest at action of mechanical loads. From the point of view of formation of perfect structure it is most rational to have a plane interface surface. However, due to specifics in heating source, conditions of heat exchange of crystal with a surrounding medium in PAZM it is impossible to obtain this front. Results of modelling of heat processes confirm this (Figure 2, c, d).

These results give an idea about changing the temperature gradient in solid phase with increase in crystal height. Here, the lower part of crystal becomes

colder and near the interface surface the temperature gradient is increased. This results in growth of temperature stresses which can cause the changes in substructure, and more clearly, the change in dislocation structure of the single crystal.

Thus, the modelling of heat processes makes it possible to study the effect of technological parameters on the formation of heat field and to predict the growth of a single crystal with a perfect substructure.

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CYLINDRICAL CATHODE-ARC PLASMA SOURCES FOR VACUUM EVAPORATION

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Problem of coating long and large-sized products in mechanical engineering necessitated development of extended plasma sources. Examples of the developed long cylindrical pulsed evaporators, their advantage and drawbacks are considered. A new long cylindrical cathode-arc plasma source with a continuous adjustable mode of arcing is proposed and its design features are described. Modifications of evaporators with a cathode length of 500 and 1550 mm are tested.

Key words: *cathode-arc evaporation, ionised vapour flow, long items, extended cylindrical evaporators, pulsed arcing mode, continuous arcing mode, controllable arc*

Vacuum coatings, produced by the methods of thermal, electron beam, cathode-arc and magnetron evaporation, are becoming ever wider applied in different fields of science and engineering. These methods in the most complete manner enable production of sound poreless coatings with a good adhesion to the product surface. Among the above methods, however, cathode-arc evaporation provides the best adhesion, in view of higher energy parameters of deposition due to the presence of an ionised vapour flow [1], which gives it an advantage over other methods of vacuum coating formation.

Cathode-arc plasma sources with a cylindrical cathode with its end face being the working surface, became the most widely applied now. Such cathode-arc plasma sources are widely used in the units for ion-plasma spraying of the type of «Bulat», «Pusk», VU-1BS, NNV-6, VUP-6, etc. [2, 3]. In these plasma sources the external electromagnetic system contains the cathode spots on the cathode end face and rotates them along circular orbits on the above surface. Such sources generate a non-uniform plasma flow of a limited size, which limits their practical application.

However, ever growing demands of mechanical engineering are raising new problems of processing extended and large-sized products, and this requires production of a plasma flow uniform over the cross-section, which cannot always be implemented by increasing the number of individual point sources of the cathode-arc plasma. In this connection, the most urgent problems in development of vacuum cathode-arc plasma sources are development of linear electric-arc sources of extended type, controlling cathode spots on the cathode working side surface and improvement of the reliability of their containment in the specified zone of cathode material erosion.

Over the last years linear sources with magnetic stabilisation of the cathode spot have been developed. In operation of an extended cylindrical pulsed evaporator the cathode spot movement is influenced by the

magnetic field of the current channel in the plasma and the current, flowing in the cathode [4, 5]. Magnetic field lines, induced in this case are tangential to the cathode surface. Cathode spots move from the firing electrode towards the current lead-in under the impact of longitudinal and azimuthal components. Thus, the cathode spots, while making a rotary motion about a cylindrical cathode, are shifted towards the current lead-in. Having reached the arc-suppressing shield, the cathode spots penetrate into a gap between the cathode and neutral shield, which is what leads to extinguishing of the discharge. The life of cathode spots on the cathode working surface is determined by the time of their displacement to the arc-suppressing shield and depends on the magnitude of discharge current.

Authors of [6] suggested the design of an extended plasma source, operating in the pulsed mode and forming a directed strip flow due to the presence of an external magnetic system. The superposed magnetic field prevents azimuthal displacement of the cathode spots due to its orthogonal component of induction, while the tangential component ensures their longitudinal displacement along a straight-line trajectory. A uniform distribution of ion current density along the cathode axis and an ordered nature of cathode spot displacement over the cathode working surface allow producing a coating, uniform along the height of the large-sized product.

For further improvement of the plasma source design and increase of the number of simultaneously processed products the authors of [7] developed a cylindrical plasma source with a radially divergent plasma flow. The operating principle of this plasma source is based on controlling the movement of cathode spots, using an external magnetic field, induced by an extended electromagnetic system, made in the form of a spiral. A negative pole of the arc power source in this case is connected via a ballast resistance to the current lead-in of the extended cathode, and the positive pole is connected to the electromagnetic system input from the side of the firing electrode, is grounded and is at the anode potential. Electromag-

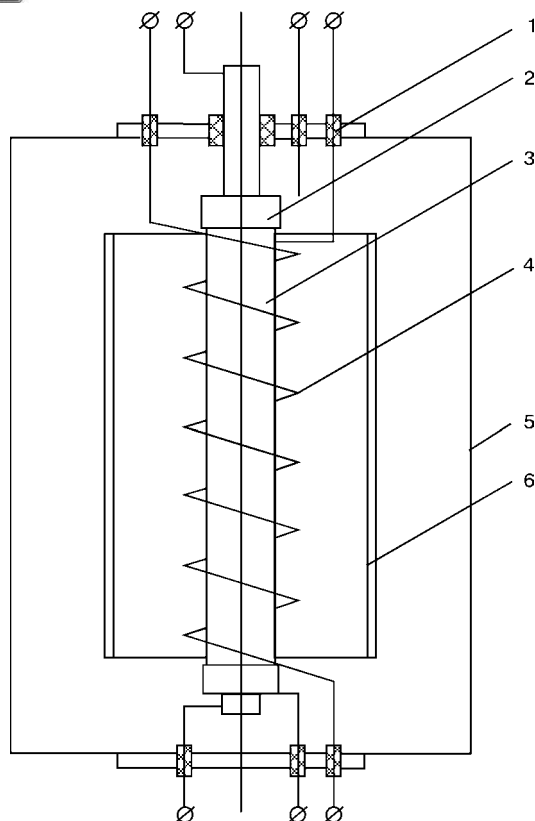


Figure 1. Schematic of an extended vacuum-arc evaporator of a cylindrical type: 1 — vacuum lead-ins; 2 — arc-limiter; 3 — cathode; 4 — anode; 5 — vacuum chamber; 6 — processed part

netic system is made of a copper tube and has water cooling.

When a pulse is applied to the firing electrode, an arc is struck, cathode spots are formed and the inter-electrode space is locally filled with the plasma of vacuum-arc discharge between the cathode and anode. The life of cathode spots on the cathode surface is determined by the time of their movement to the arc-suppressing shield. Cathode spot movement in this case is of an ordered nature, as a result of their localisation on the cathode working surface under the arch of the magnetic field, induced by the spiral magnetic system, when the arc discharge current flows in it. When the cathode spots penetrate into the gap of the arc-suppressing shield, the arc discharge is extinguished and the system goes back to the initial condition. The interval between the firing pulses is set

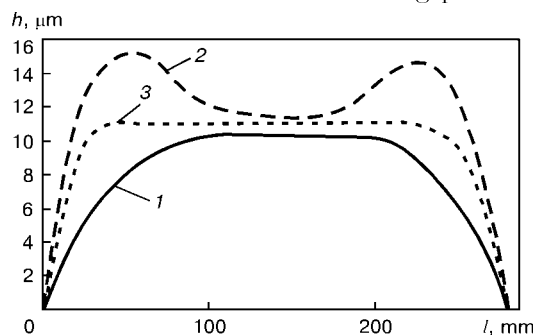


Figure 2. Relative distribution of thickness h of a coating along length l of a product: 1 — without cathode spot dwelling at the cathode ends; 2 — with cathode spot dwelling at cathode ends; 3 — optimal regulation

to be larger or equal to the mean-root-square life of cathode spots on cathode surface.

The pulsed operating mode of the evaporator and controllable nature of movement of cathode spots provide a lower working temperature of the cathode, which results in reduction of the number of drops, generated in the plasma flow and improvement of the coating quality.

Operation of such designs, however, revealed certain difficulties [5]. The pulsed mode results in much more complicated operating conditions of the electrode system of discharge initiation, which in this case should stand a greater number of starting pulses under the impact of electrothermal shock loads of a high intensity.

Operation of the device [6] further revealed that an arc-suppressing shield randomly located at the current lead-in has a significant influence on the final life of cathode spots on the working surface. Observed dwelling of cathode spots in the region of the arc-suppressing shield of up to $2-4 \tau$ (τ is their optimal time of flight) leads to a non-uniform erosion of the plasma-forming material of the cathode. It should be noted that even with a timely extinguishing of the discharge in front of the arc-suppressing shield a pit is formed on the cathode surface, which leads to the cathode spots dwelling in this zone, greater erosion of the cathode material, and, as a result, to an ineffective use of the plasma-forming material. In addition, the thickness of the formed coating on the product being processed becomes greater opposite the pit area. To eliminate this drawback, an additional shield was mounted, forming a rectilinear section of the working surface, from which erosion of the plasma-forming material proceeds.

In the design of an extended cylindrical cathode-arc plasma source, operating in the continuous mode, which is proposed by us (Figure 1), this problem is solved by adjustment of the frequency and rate of motion of the cathode spot. The operating principle of this plasma source is also based on controlling the motion of cathode spots, using an external magnetic field, induced by an extended electromagnetic system, made in the form of spiral from a copper tube. However, not arc-suppressing, but arc-limiting shields are mounted at the ends of the tubular water-cooled cathode. When the cathode spot comes closer to the arc-limiting shield, switching of the applied load occurs and the direction of the cathode spot motion is reversed. Thus, the controlled cathode spot rotates about the cylindrical cathode, continuously making a reciprocal motion with an adjustable frequency and speed.

An extended cylindrical cathode-arc plasma source, operating in the continuous mode, allows practically doubling the efficiency, compared to a similar source, operating in the pulsed mode, by eliminating the pause between the firing pulses. A uniform distribution of the ion current density along the cathode axis, an ordered nature of continuous displacement

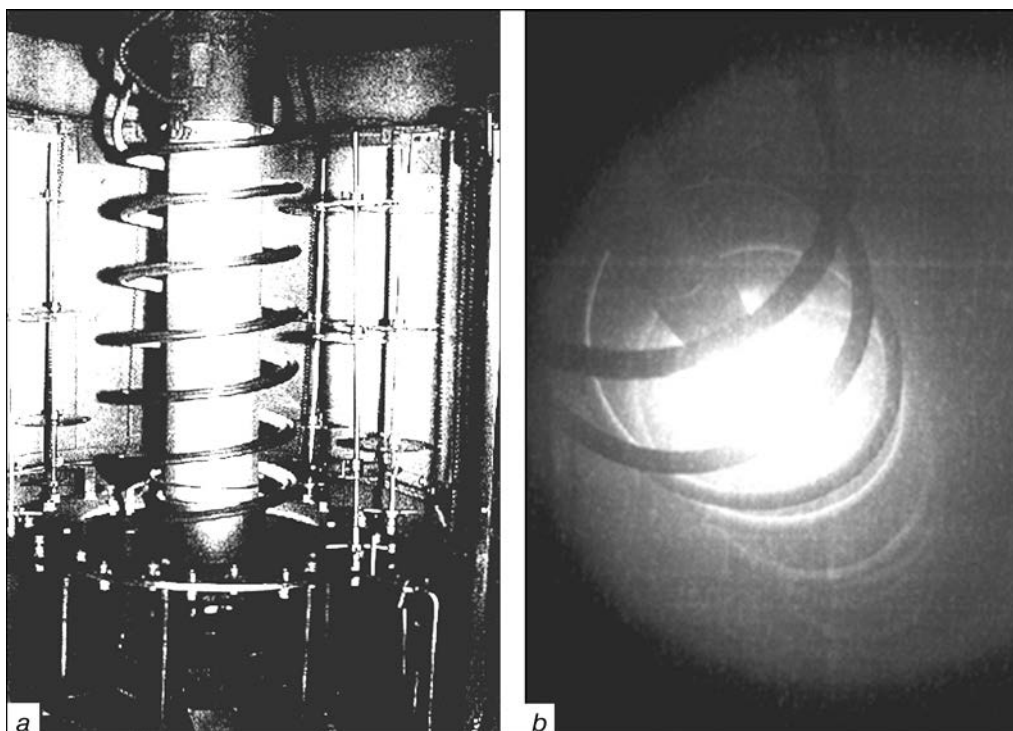


Figure 3. Extended vacuum-arc cylindrical evaporator with the side working surface of the cathode 500 mm long: *a* — appearance of evaporator inside vacuum chamber (cathode diameter of 100 mm); *b* — evaporator in operation (cathode diameter of 32 mm)

of cathode spots over the working surface of the water-cooled cathode provide a low working temperature of the cathode, which results in reduction of the number of drops, generated in the plasma flow, and improvement of the coating quality.

In the continuous arcing mode the operation of the electrode system of discharge initiation is greatly simplified, as it operates only at the initial stage of the arc firing, as well as under the conditions of its random extinguishing. In this case a more uniform consumption of the plasma-forming material of the cathode is achieved, this allowing formation of a uniform coating by thickness on an extended product. On the other hand, under the conditions of a continuous mode of the arc spot motion, the uniformity of the deposited coating thickness along the product length may be adjusted by changing the speed of the cathode spot displacement along the cathode and at its ends (Figure 2).

A cylindrical cathode-arc plasma source with a continuous arcing mode was manufactured and tested in a model variant with the cathode length of 500 mm (Figure 3). At complete loading with extended products the following values of volt-ampere characteristics were obtained: current of 200–400 A, voltage of 28–40 V, bias current of 7 A, bias voltage of 150 V and open-circuit voltage of 80 V. A new cylindrical cathode-arc plasma source with a continuous arcing

mode has been manufactured and tested now, its cathode length being 1550 mm.

Thus, a new linear cylindrical cathode-arc plasma source with a continuous arcing mode for coating deposition in vacuum has been developed, which allows applying sound coatings of a uniform thickness along the product length due to controlling the speed of cathode spot displacement about and along the axis of a water-cooled cathode, and thus enables solving many problems of mechanical engineering on processing extended and large-sized products.

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SPECIFICS OF DISTRIBUTION OF ARC VOLTAGE AND FREQUENCY OF DROPLET SHORT-CIRCUITS DURING VACUUM-ARC REMELTING OF TITANIUM ALLOYS

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The Verkhnyaya Salda Metallurgical Works conducted an experiment under industrial conditions on vacuum-arc remelting of 515 mm diameter electrode into 670 mm diameter titanium alloy ingot. Distribution of arc voltage drop and droplet short-circuiting frequency depending on the arc current, interelectrode spacing and pressure were investigated. Statistical models were constructed allowing the principle of such dependencies to be established. The models showed a clear correlation in distribution of arc voltage drops and the droplet short-circuiting frequency with variations in the interelectrode spacing. A model is suggested on the basis of which it is possible to monitor and control the melting process.

Key words: vacuum-arc remelting, titanium alloys, arc voltage drop, frequency of droplet short-circuiting, statistic models

Introduction. Main parameters of vacuum-arc remelting (VAR) control are arc current I , interelectrode spacing L and pressure P in furnace. Investigation of effect of these parameters on the melting process was carried out as long ago as the 1980s by the US scientists using linear regression models [1–4]. Among these parameters the interelectrode spacing is most difficult for control. The modern systems of VAR control use the arc voltage drop and frequency of droplet short-circuits as indices of determination of the interelectrode spacing. The perfect control of VAR process requires search and use of data containing valid information about the value of the interelectrode spacing.

The aim of the work is to study the parameters of voltage drop and frequency of droplet short-circuits and to evaluate their suitability for the VAR process monitor and control. The article is based on industrial experiment during which the effect of arc voltage drop and frequency of droplet short-circuits on change in interelectrode spacing and current in furnace was studied [5]. For each experiment the mean value of distribution of voltage, frequency of droplet short-circuiting was calculated and the statistic models from these data were developed. These models were based on independent variables. Pressure was kept constant for each experiment.

Experimental. The interelectrode spacing was changed at similar rate, i.e. 0.5 ± 0.1 mm/s. Each experiment took 10 min, then the furnace operation was stabilized for 10 min and the next measurement was started. During experiment were recorded: arc voltage drop, arc current value, frequency of droplet short-circuits and electrode movement at 2 Hz fre-

quency. All the data were reduced to a total time using industrial computer of Advantech Company. Signal of arc voltage drop was sent to the device of distinguishing the droplet short-circuitings. The latter generated pulses of 10 ms duration at short-circuitings, which was sufficient for recording separate droplet short-circuits. Modelling was performed by the method of a regression analysis which was made in IBM computer using package Statistika 5.0.

Arc voltage drop. Mathematical modelling was made using experimental results. It was decided on the basis of these results that the dependence of arc voltage drop on interelectrode spacing will be described in the range of measurements from 0 to 160 mm by a linear equation [1]

$$U_a = A + B \ln(L), \quad (1)$$

where U_a is the arc voltage drop; A and B are the coefficients of a mathematical model.

The Table gives values of coefficients for the mathematical model by arc voltage drop. Coefficients A and B were obtained from equation (1); R and D are the coefficients of regression and dispersion, respectively.

Figure 1 shows dependence of the arc voltage drop on the value of interelectrode spacing at different values of current. To plot these relationships the maximum value of regression coefficient for each value of current was used.

After determination of coefficients A and B of equation (1) their dependences on arc current were analyzed. Results are presented in Figure 2, a and b .

Consequently, knowing the arc current and arc voltage drop for each model it is possible to calculate with a certain accuracy the interelectrode spacing [2]:

$$L_{\text{calc}} = e^{(U_a - A)/B}. \quad (2)$$



Experiment	Current, A	Model by U_a				Model by F_{sh-c}				P , $\mu\text{m Hg}$
		A	B	R	D , %	A	B	R	D , %	
a1	7500	23	0.6	0.81	65.7	6.18	-0.49	0.56	32	10.3
a2		23.7	0.5	0.85	73	7.2	-0.6	0.55	30	11.72
a3		24	0.5	0.82	67	5.5	-0.4	0.55	30	13.5
b1	9000	24.6	0.5	0.79	63	4.89	-0.83	0.56	31	10.6
b2		24.4	0.6	0.81	67	5.54	-0.94	0.71	51	13.2
b3		24.19	0.7	0.89	80	3.86	-0.63	0.66	45	14.1
c1	11000	24.95	0.7	0.95	90.5	3.59	-0.82	0.75	57	9.4
c2		24.9	0.8	0.9	81	1.7	-0.45	0.61	38	11.6
c3		24.8	0.8	0.89	80	1.93	-0.52	0.66	44.4	11.9
d1	12000	25	0.9	0.93	86	3.49	-0.85	0.84	71	11.54
d2		25	0.9	0.94	88.1	3.84	-0.85	0.87	76	11.2
d3		24.3	1	0.94	84	2.33	-0.67	0.83	69.7	11
e1	13000	25.8	0.9	0.85	73	2.23	-0.45	0.72	63	10.3
e2		25.1	1	0.94	88.1	1.58	-0.64	0.85	73	9
e3		24.4	1.1	0.91	84	2.89	-0.39	0.78	62	12
f1	14000	24.95	1.2	0.96	92	1.6	-0.64	0.84	76	11.4
f2		25	1.3	0.95	90	3.32	-0.99	0.9	82	13.2
f3		23.7	1.6	0.95	91	3.43	-1	0.88	78	10.4
g1	15000	24.15	1.5	0.97	94.7	2	-0.92	0.88	78	—
g2		23.2	1.8	0.97	94.3	1.1	-0.85	0.88	78	11.3
g3		23.66	1.7	0.97	94.9	2.7	-1.11	0.8	61	10.9

Maximum error of interelectrode spacing measurement by equation (2) for all the calculated models has different values. For example, for 10 mm spacing the measurement error is ± 1.5 mm, for spacing 20 mm the error is ± 2.3 mm, for spacing 50 mm it is ± 6.5 mm, for spacing 100 mm — ± 12.7 mm. It is seen that with increase in interelectrode spacing the measurement error is also increased.

Frequency of droplet short-circuits. Frequency of droplet short-circuits F_{sh-c} depending on the interelectrode spacing is described with a sufficient precision by a non-linear equation [3]:

$$F_{sh-c} = A(L)^B, \quad (3)$$

where A and B are the coefficients from the Table for the mathematical model by frequency of droplet short-circuits. According to the Table data, with increase in arc current at unchanged interelectrode spacing the frequency of droplet short-circuits is decreased. Here, the coefficient A is also decreased. It is important to note that the coefficient of regression is increased with the current rise, i.e. quality of constructed models is growing and they become more true. Dependences of coefficients A and B of equation (3) on arc current are shown in Figure 3, *a* and *b*. Figure 4 shows dependencies of frequency of short-circuits on values of interelectrode spacing at different arc current. To plot the relationships the maximum values of regression coefficient for each value of arc current were used.

Coming from equation (3) the value of the interelectrode spacing is determined by formula

$$L_{calc} = \sqrt[B]{F_{sh-c} / A}. \quad (4)$$

Error in evaluation of the interelectrode spacing for each model is increased with a growth of its value. For example, for spacing of about 5 mm and low currents (7.5–9 kA) the measurement error is ± 1.2 mm, while for high currents (13–15 kA) it is ± 1.7 mm. For 10 mm gap and low currents the error is ± 2.9 mm, while for high currents it is ± 3.3 mm. At electrode spacing of about 20 mm and more the measurement error is ± 4 mm for currents 7.5–9 kA and ± 5.7 mm for currents 13–15 kA.

Discussion of results. In the given experiment the averaged values of arc voltage drop and frequency U_a, V

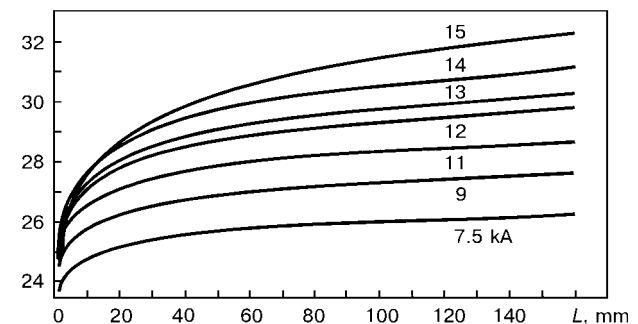


Figure 1. Predictable arc voltage drop as a function of the interelectrode spacing at different arc current values

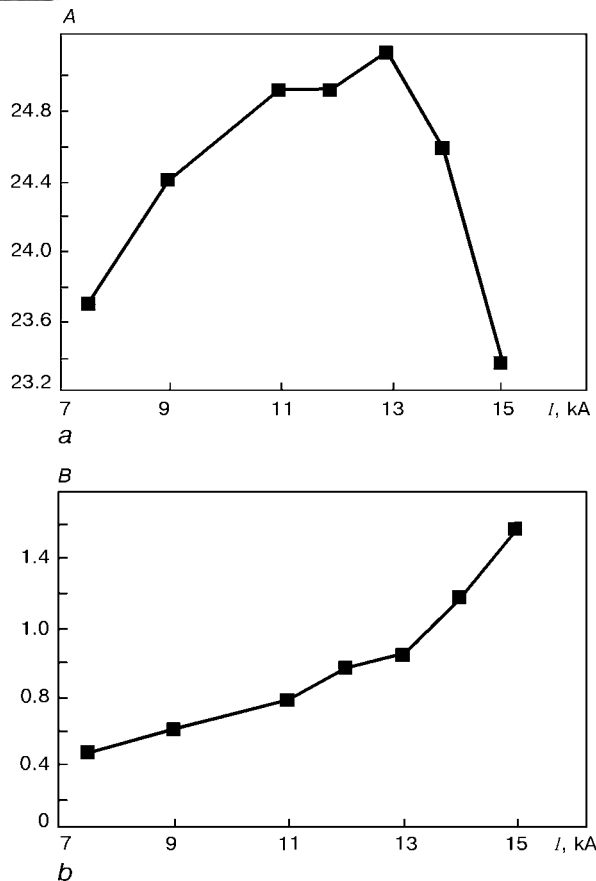


Figure 2. Dependence of coefficients A (a) and B (b) of equation (1) on arc current value

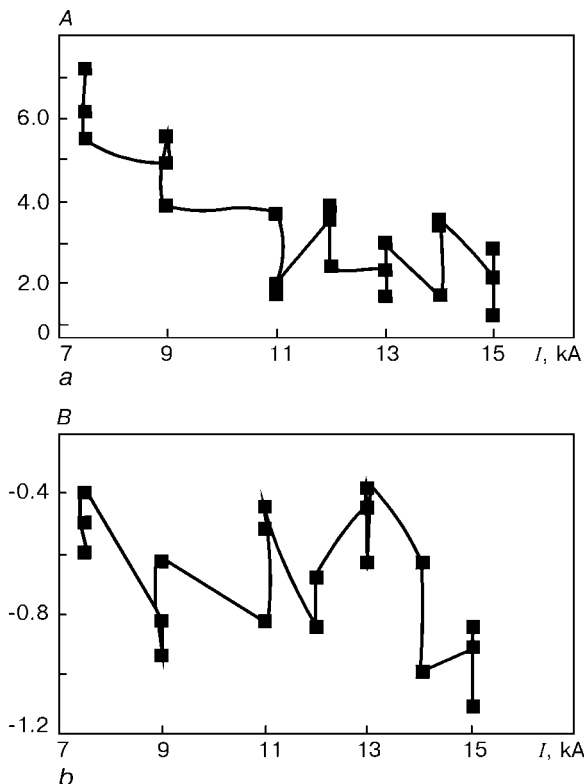


Figure 3. Dependence of coefficients A (a) and B (b) of equation (3) on arc current value

of droplet short-circuits were investigated depending on change in arc current and interelectrode spacing. Pressure in furnace was maintained in the range from 9 to 14 $\mu\text{m Hg}$. The results obtained show that the above-mentioned parameters are predictable with a change in arc current and interelectrode spacing. The constructed mathematical models do not depend on pressure and are not correlated with it in any way. Arc voltage drop is modelled with a sufficient accuracy by equation (1) depending on the interelectrode spacing.

It is known that the mean arc voltage drop reacts almost linearly on the change in the interelectrode spacing. Many controllers of vacuum arc remelting were based on this phenomenon. In our investigations the non-linear models of calculation of the interelectrode spacing were constructed by variation in arc voltage.

Equation (1) is none other than Ohm's law: the voltage drop is growing with increase in distance between anode and cathode, i.e. interelectrode spacing plays a role of «resistance». Coefficient A , the first term of equation (1), is constant and its value does not depend on variation of the interelectrode spacing. Coefficient A is growing linearly with increase in current value from 7.5 to 13 kA and decreased at further current increase that is illustrated in Figure 2, a. This fact was caused, probably, by the effect of magnetic field generated directly by the electric arc and influencing the movement and formation of cathode spots. The voltage drop occurs, as is known, in the near-cathode region and depends on density and amount of cathode spots which play a role of parallel-connected conductors of resistances.

Coefficient B (Figure 2, b) is increased proportionally with a growth of current value. This proves that the share of voltage drop between anode and cathode, brought in by the second term of equation (1) with respect to the total arc voltage drop, is increased depending on the arc current value.

The change in frequency of droplet short-circuits by value of interelectrode spacing depends on the arc current value. Figure 4 shows models of changing frequency of droplet short-circuits depending on the interelectrode spacing for each current value. The higher arc current value, the lower frequency of drop-

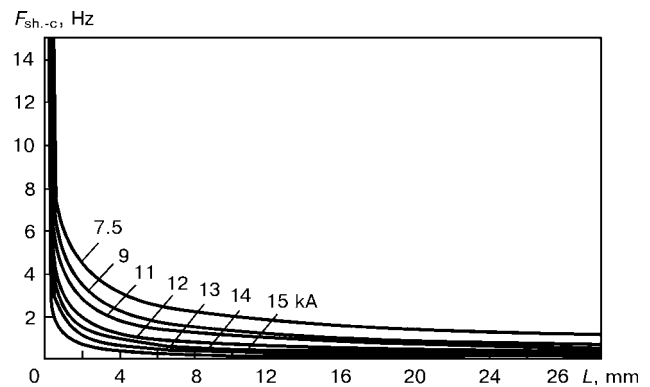


Figure 4. Predictable frequency of droplet short-circuiting as a function of the interelectrode spacing at different arc current values



let short-circuits at unchanged value of the interelectrode spacing. Coming from this it is possible to conclude that with a growth in arc current value the frequency of dropping is decreased.

For each constructed mathematical model a maximum deviation ΔU_a and $\Delta F_{sh.-c}$ from real value, which leads to error in measurement of the interelectrode spacing, was determined. The error is increased with a growth of the interelectrode spacing.

The transition process should be outlined specially. It occurs in changing a diffusive discharge for a constricted discharge and vice versa and causes the jump in voltage drop and increase in frequency of droplet short-circuits in the interelectrode spacing of 25–40 mm. This is especially typical of low current 7.5 kA. At high currents of order 14–15 kA this transition process is not observed and parameters of voltage drop and frequency of droplet short-circuits behave themselves «calmly», not changing much. Models, constructed for high currents, are more true that is confirmed by values of coefficients of regression and dispersion from the Table.

CONCLUSION

Results of a multifactor experiment carried out at Verkhnyaya Salda Metallurgical Works under the industrial conditions give the following conclusions:

1. Arc voltage drop has a non-linear dependence on interelectrode spacing and defined by none other than Ohm's law. In this case the interelectrode spacing

plays the role of resistance. Accuracy of a predictable spacing for each model is ± 1.5 mm at 10 mm and decreased with increase in spacing.

2. Frequency of droplet short-circuits has a non-linear dependence on the interelectrode spacing and decreased with its increase. The accuracy in determination of spacing value for each model is decreased with increase in arc current value.

3. Transition processes occurring in changing the type of discharge, influence the parameters of voltage drop and frequency of droplet short-circuits, thus leading to a non-adequate evaluation of the interelectrode spacing.

4. The obtained models of changing voltage and frequency of droplet short-circuits depending on the interelectrode spacing can be used with a sufficient accuracy for determination of the latter in the process of a vacuum-arc remelting of titanium alloys and high-alloyed steels.

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MATHEMATICAL SIMULATION OF THE MODES OF CAST IRON HEATING AND SOAKING IN A MAGNETODYNAMIC HOLDING FURNACE

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Investigation results are given on thermal modes of operation of a magnetodynamic holding furnace for heating and pouring of cast iron. A mathematical model of the process under investigation was developed. Integrated computational experiment and full-scale tests in operating equipment were conducted. Factors, determining occurrence of thermal processes in the furnace, were established. Process parameters were established, allowing optimisation of the conditions of heating and holding the cast iron in a magnetodynamic holding furnace.

Key words: magnetodynamic holding furnace, mathematical model, temperature, cast iron heating, holding, power, metal motion rate

Thermal characteristics of electrotechnical casting installations in the mode of holding, heating and pouring of the melt are highly important for production, as they determine power losses for preparation and pouring of metal, its holding during the non-working shifts, as well as in the pauses between the operations of the technological cycle of casting. Lowering the power consumption of these installations in casting production is a task of primary importance to reduce the cost of products and improve its competitiveness. This is particularly true for equipment, used for melting and mixing of metals, in particular, induction channel furnaces and electrotechnical installations for cast iron holding, overheating and pouring [1], including such as magnetodynamic holding furnaces [2].

A feature of channel-type electrotechnical installations consists in that heating of the main mass of metal in the crucible is due chiefly to the overheated melt flow, coming from the channel [3]. Therefore, dynamics of metal heating in the crucible depends both on the holding furnace design, applied thermal insulation and refractory materials, and on the electric power and hydrodynamics of melt flow in the channel of such an installation.

Analysis of operation of a magnetodynamic holding furnace for cast iron pouring shows that it, as a rule, contains a considerable mass of the metal, which during castings production varies in a broad range, which may lead to a change of the melt pouring temperature. Therefore, features of metal heating and holding should be taken into account, when developing the technology to provide a stable thermal condition of the melt.

Thermal processes in a magnetodynamic holding furnace in production of cast iron castings were studied, using the developed mathematical model. Model adequacy to the actual processes was confirmed by

earlier conducted investigations under the conditions of an operating foundry [4]. When a computational experiment was conducted, process parameters were varied in the following ranges: inductor power $P_{\text{ind}} = 0-1000$ kW, cast iron weight in the crucible $M = 1000-4000$ kg, speed of melt flow in the induction channel $v_{\text{mov}} = 0.01-0.5$ m/s. Intensity of cast iron heating in magnetodynamic holding furnace depends, primarily, on inductor power and melt weight. Treatment by regression analysis methods of the obtained results on cast iron overheating in the holding furnace up to temperatures in excess of 1400°C , yielded an analytical dependence of variation of metal temperature T in time on inductor power and melt mass in the holding furnace at a fixed rate of transit flow of the metal in the channel of 0.05 m/s:

$$T = 1412.6119 - 2.4867 \cdot 10^{-7} \tau^2 + 0.0004 \tau P_{\text{ind}} - 6.7896 \cdot 10^{-6} \tau M - 1.6943 \cdot 10^{-8} \tau P_{\text{ind}} M, \quad (1)$$

where τ is the time of cast iron heating at a specified electric power of inductor, s. Correlation factor is 0.9391 .

Having differentiated expression (1) with respect to time, we obtain an expression for evaluation of heating rate ($^\circ\text{C}/\text{s}$) of the metal in the holding furnace crucible:

$$\frac{\partial T}{\partial \tau} = -4.9734 \cdot 10^{-7} \tau + 0.0004 P_{\text{ind}} - 6.7896 \cdot 10^{-6} M - 1.6943 \cdot 10^{-8} P_{\text{ind}} M. \quad (2)$$

Solving equation (2) on condition of the rate of cast iron heating being equal to zero, allowed establishing such a relationship of the melt mass in the holding furnace and the inductor power, which enables maintaining the specified volume of cast iron for the required time at a fixed temperature with minimal power losses. Results of the computational experiment are given in Figure 1 and are interpreted as follows. For instance, minimal mass of the melt,



which may be held in the furnace at the temperature of 1400 °C, depending on inductor power, may be determined from expression (2):

$$M = \frac{0.0004 P_{\text{ind}}}{6.7896 \cdot 10^{-6} + 1.6943 \cdot 10^{-8} P_{\text{ind}}} \quad (3)$$

Curves 1–5 in the above Figure correspond to the operation of cast iron holding in magnetodynamic holding furnace at melt temperatures in the range of 1250–1550 °C. The region under each curve corresponds to the mode of metal heating above the specified temperature, and the region above any curve corresponds to the mode, which is inadmissible under the conditions of production, as in this case the melt is cooled below the specified holding temperature.

An integrated computational experiment allowed determination of the total heat losses in a magnetodynamic holding furnace of MD-4000 model at different mass and temperature of cast iron, held in it, and thus evaluation of the power consumption in the metal holding modes [5]. It was established that with the increase of the melt mass from 1500 up to 4000 kg, and increase of its temperature from 1250 up to 1550 °C, heat losses rise more than 3.5 times — from 37 up to 136 kW. According to the obtained data, in order to save power, it is recommended to hold cast iron in the holding furnace at a minimal admissible temperature of 1250 °C. Proceeding from the mass of the melt in the holding furnace, inductor power in the holding mode may be chosen, sufficient for compensation of heat losses at a given temperature.

The second factor in importance, which is responsible for the thermal condition of cast iron in the magnetodynamic holding furnace, is the intensity of heat exchange between the zone of the main heat evolution — induction channel of the furnace — and the crucible, which may hold up to 95 % of useful mass of the melt, contained in the furnace. In practical work such a mode of metal moving in the channel–crucible system is sought that the following recurrent factors are taken care of:

- at a low rate of metal motion high overheating of cast iron in the channel occurs, causing a premature failure of the holding furnace induction part, because of lining wear;
- at a high rate of melt circulation between the channel and crucible and concurrent considerable power of heat released in the channel, the entire volume of metal in the holding furnace has a high temperature, this leading to higher heat losses and increased power consumption, accordingly, greater loss of the main and alloying elements of the alloy, premature wear of the crucible lining and in the general case, higher product cost.

In order to find the methods for optimisation of such heating, a computational experiment was performed to evaluate the influence of the speed of cast iron motion in the holding furnace channel on the thermal condition of the metal. It is found that at cast iron flow rate in the channel of 0.01 m/s the

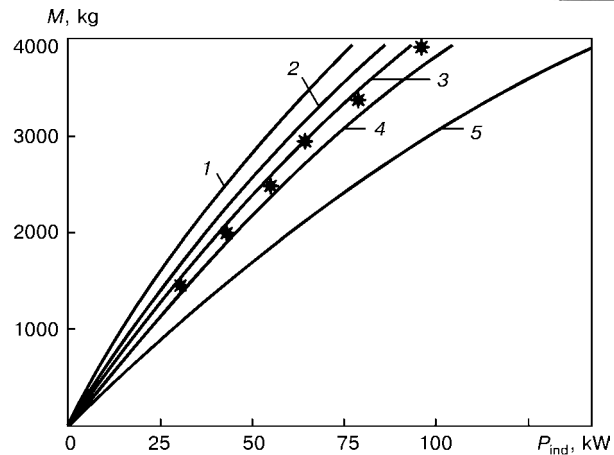


Figure 1. Relationship of the mass of cast iron in a magnetodynamic holding furnace and inductor power: 1–5 — results of computational experiment (holding at the temperature of 1250 (1), 1350 (2), 1400 (3), 1450 (4), 1550 (5) °C); * — data of full-scale experiment, obtained at the temperature of cast iron holding of 1400 °C

melt is in the zone of the main heat evolution for a long time (more than 5 min), because of a considerable length of the channel (about 2.7 m) and becomes considerably overheated. On the other hand, a much smaller amount of heat is applied to the main mass of the metal, which is in the mixer crucible, than at intensive circulation of cast iron in the channel–crucible system.

As follows from the derived results, at the initial moment of time, the metal in the furnace crucible is held at the initial temperature and only after some time, with the arrival of highly overheated melt from the channel, the cast iron temperature in the crucible begins rising (Figure 2, curves 1, 4).

The high speed of metal motion of 0.5 m/s (curves 2, 3) leads to the melt being in the induction channel for just 6 to 7 s, being only slightly heated, and the heat absorbed by the melt being immediately and almost completely transferred to the cast iron, contained in the installation crucible. As a result, the entire volume of cast iron, both in the channel and the furnace crucible is at the same temperature.

Quasistationary mode is characterised by a steady-state linear rise of cast iron temperature and constant level of metal overheating in the channel–crucible system, and such a mode is reached by the process at

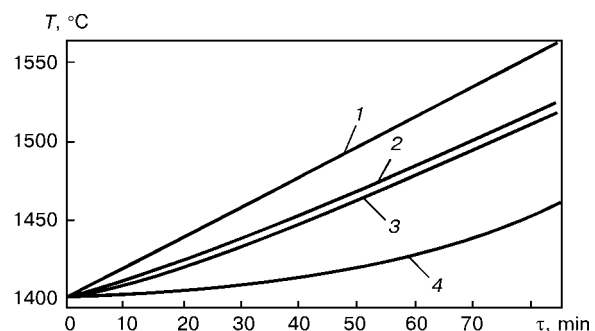


Figure 2. Variation of cast iron temperature in a magnetodynamic holding furnace at different speeds of metal motion ($M = 1500$ kg, $P_{\text{ind}} = 100$ kW): 1, 4 — in the channel and the crucible, at $v_{\text{mov}} = 0.01$ m/s, respectively; 2, 3 — same at $v_{\text{mov}} = 0.5$ m/s



$v_{\text{mov}} = 0.01 \text{ m/s}$ after $\approx 80 \text{ min}$, and at $v_{\text{mov}} = 0.5 \text{ m/s}$ after $\approx 20 \text{ min}$.

In addition, different angles of inclination of the rectilinear sections in Figure 2 indicate that for a quasistationary mode the relative rate of cast iron heating at a low speed of melt motion is higher than at a high rate.

Processing of the computational experiment results by the methods of regression analysis for a fixed mass of cast iron of 4000 kg in the installation yielded the following dependence of variation of overheating temperature:

$$T = 1427.9673 - 0.0209 \tau - 4.3426 \tau^2 - 0.3430 P_{\text{ind}} + \\ + 0.001119 P_{\text{ind}}^2 + 0.0003 \tau P_{\text{ind}} - 57.1965 v_{\text{mov}} + \\ + 89.6303 v_{\text{mov}}^2 - 0.37 \cdot 10^{-4} \tau v_{\text{mov}} P_{\text{ind}}. \quad (4)$$

Correlation factor was 0.9976.

Experimental verification of the results of mathematical simulation was conducted in an operating foundry, when producing cast iron castings with application of a magnetodynamic holding furnace. The power of the holding furnace inductor was varied in the range of 100–300 kW. Electric parameters of the furnace were changed by remote control from the panel, using autotransformers AOESK-180/420, mounted in electric power circuits of both the inductors.

Thermal condition of cast iron was controlled using a continuous temperature monitoring unit, designed by Physical-Technological Institute of Metals and Alloys of the NAS of Ukraine [6]. The main element of the unit is the light guide. In addition, thermocouples of PR 30/6 and VR 5/20 type with disposable metering nozzles and millivoltmeters as indicating instruments were also used for these purposes.

Thermal condition of cast iron was studied in its holding in different zones of the crucible and the channel of magnetodynamic holding furnace by direct measurement of temperature in fixed points. It was established that temperature gradient in the crucible and the channel (in the working zone at the furnace pouring nozzle) is not higher than 20°C . This confirms the presence of intensive heat exchange between the channel and the crucible of the furnace, due to the impact of electromagnetic forces in the channel working zone. Knowing the geometrical dimensions of the channel, the speed of melt motion in the induction device channel may be calculated by the extent of metal overheating [7]:

$$v_{\text{mov}} = \frac{\eta_{\text{el}} \eta_{\text{th}} P_{\text{ind}}}{\Delta T S c \rho}, \quad (5)$$

where η_{el} , η_{th} are the electric and thermal efficiency of the device (in this case, magnetodynamic holding furnace); ΔT is the melt temperature difference between two points along channel length, $^\circ\text{C}$; S is the cross-sectional area of the induction channel, m^2 ; c is

the specific heat capacity of the metal, $\text{J}/(\text{kg} \cdot ^\circ\text{C})$; ρ is the melt density, kg/m^3 .

Using experimental data obtained during full-scale investigations and dependence (5), it was calculated that the speeds of melt motion in the magnetodynamic holding furnace can be in the range of 0.01–1 m/s, when cast iron circulation in the channel–crucible system is provided.

The adequacy of the mathematical model, developed and used in the computational experiment, was confirmed during experimental determination of the rational modes of cast iron holding in the magnetodynamic holding furnace. So, for instance, at the holding temperature of 1400°C the dependence between the cast iron power and mass in such a holding furnace, derived in operating equipment, differs by not more than 5 % from the calculated dependence (see Figure 1).

The results obtained during the computational and full-scale experiments were used for development of advanced power-saving technology of induction overheating of cast iron in its electromagnetic pouring with application of a magnetodynamic holding furnace [5, 8]. By the proposed technology it is recommended to hold the melt at a low temperature and overheat it to the pouring temperature during discharge from the holding furnace. Practical application of the new technology will allow:

- reducing the power consumption during holding, overheating and pouring of cast iron by 5–8 kW·h/t;
- reducing the loss of the main and alloying components by 2–5 % due to metal holding in the furnace at a lower temperature;
- extending the service life of lining of magnetodynamic installation and reducing the cost of refractory materials by 10–15 % due to lowering of thermal loads on the working layer of lining;
- computerising and automation of the modes of holding and differentiated heating of cast iron for a broad range of castings, made from initial metal of the same composition.

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ABOUT BEHAVIOUR OF NITROGEN IN GAS-SLAG-METAL SYSTEM

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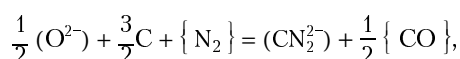
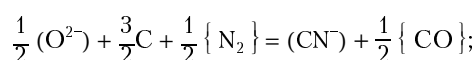
Analysis of behaviour of nitrogen in the gas-slag-metal ternary system was made on the basis of review of literature data. Interrelation of nitrogen solubility and transport in slags with their structure is considered. It has been established that metal can be nitrided directly from the gas phase through oxide-fluoride slags.

Key words: nitriding, oxide-fluoride slags, solubility, oxidizing potential

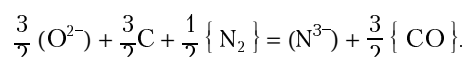
Demand for high-nitrogen steels during recent years made the researchers-metallurgists to face the problem of metal nitriding directly from gas phase during the process of its melting.

For plasma-arc remelting, which provides a contact of gas activated in plasma with a molten metal, this problem was solved to a certain extent due to fundamental studies of binary gas-metal system [1, 2]. However, in parallel with plasma-arc remelting there are other methods of remelting in which special slags are used as an obligatory element. It is naturally, that the presence of a slag pool over the metal melt changes significantly the process of absorption of nitrogen with metal from the gas phase and, therefore, a thorough study of nitrogen behaviour in ternary gas-slag-metal system is required.

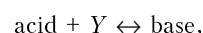
A bank of publications devoted to the nitrogen behaviour in the ternary system consists of several tens of works, some of them are given in reference at the end of article. The analysis of these works shows that, firstly, the process of nitrogen dissolution from the gas in a slag melt is governed by a law of a square root [3-14] and, secondly, there is no single opinion among the researches in description of a form of nitrogen existence in liquid oxide and oxide-fluoride melts. According to some data the nitrogen in slags is available in the form of ion N^{3-} [3, 7, 13, 14], while from other data it is in the form of ions of cyanides CN^- and cyanamides CN_2^{2-} [5, 6], though there is an opinion about a simultaneous existence of either ions [4, 8-11]. The differences in opinions are based on different interpretation of the role of carbon available either in metal or in gas, or in a crucible material. If carbon comes into reaction with nitrogen, then cyanides or cyanamides are formed:



and, if carbon plays a role of a slag deoxidizer, then nitrogen exists in it in the form of ion N^{3-} :



But despite of the difference in forms of nitrogen existence in the slag, all the researchers outline the dependence of its content on a partial pressure of oxygen [7, 12, 13, 15, 16] or CO [11] in a gas phase. In accordance with the theory of media with covalence electrons [17], based on the Luxe's redox theory [18], well-known to chemists and forgotten sometimes by metallurgists, any interaction in a condensed phase, containing a strong oxidizer, is governed by the relationship



where Y is the negatively charged particle (electron, anions O^{2-} , F^- , etc.).

If to keep a partial pressure of impurity component (nitrogen, sulphur, etc.) over the oxide phase constant and to vary an oxidizing potential in system, then, in spite of unchanged activity of the impurity, its solubility can change by several orders [17]:

$$c_i = k_i a_{O_2}^{\frac{v}{2}},$$

where v is the valance of impurity in slag with appropriate sign (nitrogen in slags has a valance 3-); k is the coefficient which depends on temperature, slag composition and nature of impurity; a_O is the oxygen activity in slag. As to the nitrogen solubility in slags, the law of square root is expressed as [15, 17]:

$$\lg(N) = k p_N^{\frac{1}{2}} p_{O_2}^{-\frac{3}{4}},$$

where p_N , p_{O_2} are the partial pressure of nitrogen and oxygen, respectively. Therefore, in study of nitrogen solubility in slags, based on oxides, it is necessary to take into account the presence of active oxygen both in slag and slag phase.

In our opinion, the role of carbon in the processes considered is the following. Entering the oxide melt the carbon deoxidizes it, i.e. it interacts with active oxygen with CO formation. Only after the completion of deoxidation the carbon interacts with slag cations



forming compounds close to carbides by composition, and they react in turn with nitrogen from the gas atmosphere forming cyanides or cyanamides. This was confirmed by investigations of solubility of nitrogen in pure calcium fluoride melted in a graphite crucible and bubbling with a gaseous nitrogen [19].

In study of nitriding the researchers study not so much the nitrogen solubility as its transfer through the slag to the molten metal. This is a diffusion process and characterized by a low rate, i.e. the diffusion coefficient for oxide slags of $\text{CaO-SiO}_2\text{-Al}_2\text{O}_3$ system is $(2-5) \cdot 10^{-6} \text{ cm}^2/\text{s}$ [20, 21]. As is known, the coefficient of diffusion of element is inversely proportional to the medium viscosity, in which the element is diffused. Therefore, it is possible to accelerate a little the nitrogen transfer by slag dilution (for example, by adding CaF_2). It is possible also to control the nitrogen transfer across the slag thickness and by changing the oxidizing potential at the interface of phases [22]. If the oxidizing potential at the slag-metal interface is lower than that at the gas-slag interface, then the nitrogen will tend to the metal.

Researchers did not almost deal with the study of nitrogen transition into metal from slag, though a difference of potentials can be created at interface of these phases to prevent the nitrogen transition into metal. Therefore, the transition of nitrogen from slag into metal, accompanied, probably, by electron emission, should proceed simultaneously with transition of any other elements from metal to slag with an absorption of the same or larger amount of electrons.

Over the recent years many works appeared devoted to structure of oxide melts. Earlier, the researchers studied the solidified slag and made assumptions about its structure in the molten state. The use of spectral methods as regards to slag melts makes it possible to study their real structure. Therefore, it is necessary to dwell more in detail on the modern conceptions about the slag melts.

The oxide slags are the packing of oxygen anions, which is close to the most dense packing, where basic and acid cations are clustered in octa- and tetravoids, respectively [7, 23, 24, 25]. This system is capable to be reconstructed very quickly that is caused by the processes of interaction both in the slag itself and also with contacting phases, such as gas, metal and crucible material. At such reconstructions the amount of active oxygen, available in the space between the clusters of complex compounds, and (or) the occupation of bonds Me-O and O-O . Here, oxygen can exist in atomic or ionized state [24, 25]. This leads to the change in oxygen activity, so to the change in oxidizing potential. Consequently, at the slag interfaces an oxidizing potential is established which is typical of each of the contacting phases. This potential will define the interaction of slag components with those of contacting phases [22].

The role of calcium fluoride in oxide melts is also interesting. Welding experts consider that calcium fluoride is added to slag to create a protective atmosphere over the slag melt consisting of volatile fluorides, for example, SiF_4 . Metallurgists consider calcium fluoride

as a diluter of slag melts. However, nobody considers it as an active participant in the processes of interaction, though there are data about metal nitriding in ESR under the oxyfluoride slags [8, 19].

Thus, the analysis of literature data makes it possible to conclude that, firstly, the metal nitriding directly from the gas phase through oxide and oxyfluoride slags is possible, and, secondly, it is necessary to define conditions influencing the rate of metal nitriding through the slag and to master the control of this process.

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IDENTIFICATION OF TYPES OF OXIDE INCLUSIONS IN STEEL 110G13L BY FRACTIONAL GAS ANALYSIS EVALOGRAMS*

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The content of oxygen combined into different types of non-metallic inclusions in commercially melted steel 110G13L samples was evaluated by the method of fractional gas analysis using the ingenious software OxSeP and the LECO gas analyser TS-436. It has been confirmed that the oxygen content of inclusions and their types depend upon the residual content of a deoxidising element (aluminium) in steel.

Key words: Hadfield steel, melting technology, mechanical properties, fractional gas analysis, oxide inclusions, nitrogen

Increasing requirements imposed by customers on the quality of metal products make it necessary to improve existing technical solutions and look for the new ones from the scientifically grounded standpoints. This problem is especially pressing for high-manganese austenitic steel 110G13L (wt. %: 11.5–16.5Mn; 1.0–1.3C; 0.3–0.9Si; $\leq 0.02S$) (GOST 7370–86), the castings of which are widely applied in transport engineering to manufacture cores and all-cast frogs of switches of trunk railways, parts of mining and agricultural machinery and other critical products [1].

Castings of steel 110G13L are made, as a rule, at numerous foundries of machine building factories and metallurgical plants using the following process flow diagram: melting of steel in basic arc furnaces using raw charge or in-house wastes, pouring of metal into sand-loam moulds (or permanent moulds) and subsequent heat treatment of castings [2].

At one of the factories steel 110G13L is made in basic electric furnaces of the DSN-6 type using the charge consisting of the carbon steel scrap. The concentration of carbon after melting is 0.4–0.6 %. Oxidation of phosphorus and carbon is done using iron ore. The oxidation period is ended upon reaching the 0.1 % P and 0.15–0.2 % C content of the steel, the oxidation rate being 0.35–0.6 % C per hour. Slag is removed at the end of the oxidation period to prevent a reverse transfer of phosphorus from slag to metal during alloying the steel with manganese. Lumpy ferrosilicium and ferromanganese are fed to the metal surface in such amounts that the steel will contain 0.6 % Si and 1.5 % Mn. Then the reducing slag with a basicity of 1.6–2.0, consisting of lime, fluorspar, ground coke and finely crushed ferrosilicium, is in-

duced. After deoxidation the metal is alloyed with ferromanganese FMn78. Final deoxidation of the steel is performed using aluminium pigs in the furnace and ladle, the aluminium consumption being up to 5 kg per ton of the steel. Cast cores and frogs are produced in sand-loam moulds. The cores and all-cast frogs are subjected to heat treatment under the following conditions: heating to 1000 °C, isothermal holding for 5–6 h at a temperature of 1000–1100 °C and water quenching.

Study of structure of the cast steel 110G13L, consisting of high-manganese austenite, showed that it contained oxide inclusions [3] and precipitates of the carbon phosphide phases, the morphology and amount of which are determined in many respects by the chemistry of the steel, solidification conditions, cooling of castings and their heat treatment parameters [1].

It should be emphasised that some defects revealed in castings of this steel (contamination with endogenic non-metallic inclusions, increased gas content, sieve-type porosity, hot and cold cracks) are associated mostly with the technology used in the oxidation and reduction periods of melting, parameters of deoxidation of steel in the furnace and ladle and casting parameters, whereas heat treatment fails to remove them. Therefore, the high-quality cast cores and frogs can be provided only by keeping to the scientifically grounded parameters of each stage of their start-to-finish complicated production technology.

Mechanical properties of the cast cores and frogs of steel 110G13L for trunk railways should meet requirements of GOST 7370–98 (Table 1).

Statistical processing of a large array of data on production of railway switch cores and frogs from steel 110G13L showed [4, 5] that with the earlier steel making technology the yield of the first group of products did not exceed 15–20 %, which led to

*The term «evalogram» originates from an English word «evaluate» (gas).

**Table 1.** Mechanical properties of steel 110G13L castings

Group of castings	Tensile strength, kgf/mm ²	Yield stress, kgf/mm ²	Elongation, %	Reduction in area, %	Impact toughness KCU, kgf-m/cm ²
I	>90.1	>36	>27.1	>30.1	>25.1
II	80.1–90	>36	22–27	25.1–30	20.1–25
III	70–80	>36	16–25	16–25	17–20

decrease in cost effectiveness and competitiveness of these metal products in the national and foreign markets of metallurgical and machine building industries.

In view that quality of the cores and frogs of a high-manganese steel depends upon a number of technology factors, the purpose of the study described in this article was to evaluate the content of oxygen and nitrogen, as well as the types (phase composition) of oxide inclusions in castings of different quality groups.

Investigation objects and procedure.* The gas (oxygen and nitrogen) content was determined on metal samples of 10 commercial heats subjected to mechanical tests (Tables 2 and 3). According to the standards in force, specimens for mechanical tests were cut from a technology sample, i.e. bar (Figure 1), after its heat treatment together with a batch of the cast cores and frogs.

The amount of oxygen adsorbed by the sample surface, dissolved in metal and combined with it to form different types of oxide non-metallic inclusions was determined by the fractional gas analysis (FGA) method developed and used to advantage for investigations by the Diagnostics Laboratory of the A.A. Bajkov Institute for Metallurgy and Materials Science of the Russian Academy of Sciences [6, 7].

The FGA method is a modification of reduction melting of steel samples in the graphite crucible in a carrier gas flow at a preset heating rate. Analytical

procedure of the FGA method involving the LECO analyser TS-436, allowing fractionation to be conducted using the specially developed ingenious software, includes the following operations: analysis of a sample by the optimal program in terms of resolution of the oxide spectrum, processing of the analysis results using the OxSeP software from the analyser data buffer and identification of the precipitated oxide types.

Data processing is performed automatically with no involvement of an operator, and includes the following stages:

- blank procedure for dummy correction of non-isothermal analysis, i.e. accounting for the effect of temperature-dependent functions of gas evolution from an empty crucible during the heating process, and effect of active sublimates and impurities in the carrier gas during the non-isothermal analysis;
- procedure for reconstruction of the initial curve of gas evolution from the analysis results. The procedure eliminates the effect of dilution of signal of carbon oxide in the carrier gas, effect of gas evolution of the system in movement of the analytical gas from the graphite crucible of the furnace to the IR-cell of the analyser;
- procedure for separation of the analytical curve into the spectrum of peaks of oxide gas evolution, based on solution of the inverse problem using the developed physical-chemical model of the process.

The standard sample of the Iron and Steel Institute of Japan (JSS-C-S-1c), having the following metrological characteristics, was used for calibration of the instrument: certified value of the mass fraction of oxygen 0.004 % and that of nitrogen 0.0254 %, at the values of an absolute error for oxygen equal to 0.00038 % and nitrogen 0.0012 %. In addition, the use was also made of the LECO standard calibration steel sample having the following metrological characteristics: certified values of mass fraction of oxygen

Table 2. Chemistry of the commercially melted steel 110G13L

Heat No.	Content of elements, wt. %					
	Mn	C	Si	P	S	Al
2-121	11.5	1.10	0.48	0.068	0.011	0.017–0.022
2-167	13.1	1.10	0.30	0.079	0.013	0.015–0.040
2-156	13.6	1.00	0.70	0.076	0.012	0.090–0.006
2-133	16.5	1.16	0.48	0.078	0.011	0.022–0.006
2-150	12.6	1.20	0.64	0.089	0.012	0.006–0.022
2-152	14.2	1.12	0.30	0.075	0.013	0.017–0.035
2-153	12.8	1.21	0.62	0.090	0.011	0.035–0.015
2-155	13.5	1.08	0.54	0.089	0.012	0.015–0.055
2-139	14.8	1.16	0.52	0.068	0.012	0.005–0.035
2-96	13.5	1.08	0.52	0.077	0.013	0.016–0.020

*L.S. Kvasova participated in the investigations.

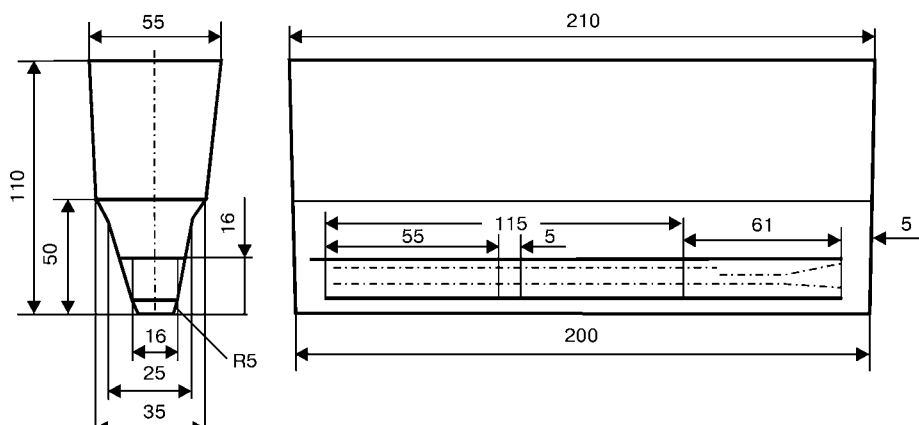


Figure 1. Experimental bar for evaluation of mechanical properties of steel 110G13L.

0.0111 % and nitrogen 0.0071 %. The absolute error for oxygen was 0.0002 % and for nitrogen — 0.0003 %.

The oxygen and nitrogen contents of the samples studied were determined by the method of reduction melting of steel samples 0.3–0.7 g in weight in the inert gas (helium) atmosphere under the following conditions: time of degassing the crucible 20 s, temperature of degassing the crucible 2400 °C, minimum time of analysis of oxygen 30 s, that of nitrogen — 40 s, analysis temperature 2200 °C and analysis time 400 s.

Experimental results and their analysis. Table 4 gives data on the content of different types of oxygen contained in steel samples: total content, including surface oxygen and dummy correction oxygen $[O]_1$, oxygen content of the sample, including surface oxygen less the dummy correction oxygen $[O]_2$, dummy correction for the oxygen content $[O]_3$, amount of oxygen adsorbed by the sample surface $[O]_4$ and oxygen content of inclusions $[O]_5$.

It follows from this Table that, despite keeping to the existing technology for melting and loam-mould casting of steel 110G13L, the total oxygen content of the investigated samples of different heats varies within wide ranges from a minimum value of $50.6 \cdot 10^{-4}$ to a maximum of $532 \cdot 10^{-4}$ wt.%. In this case the major amount of oxygen in the samples is contained in oxide inclusions: from $32.7 \cdot 10^{-4}$ to $502 \cdot 10^{-4}$ wt.%, which is

caused by both differences in residual concentration of aluminium and endo- and exogenic character of their origin.

Of scientific and practical interest are the results of fractional gas analysis of the steel samples in terms of evaluation of different types of non-metallic oxide inclusions using the OxSeP software.

Given below is the analysis of evalograms (Figures 2–8) of gas evolution in reduction melting of a sample and separation of the analytical curve into the spectrum of evolution of gases of different types of oxides, based on the procedure for solution of the inverse problem by the developed physical-chemical model of the process. Evalograms are given in the order of an increasing content of oxygen combined to form inclusions. Identification of peaks of the inclusions was performed by parameter T_b using the special OXID software.

Sample 2-150 (Figure 2). $[O]_5 = 32.7 \cdot 10^{-4}$ wt.%. The following types of inclusions were identified from the evalogram: (Mn, Fe)O, SiO_2 , $MnO \cdot Al_2O_3$ and $Al_2O_3 \cdot SiO_2$. Temperatures (T_b , K) were: $2FeO \cdot SiO_2$ — 1349, MnO — 1592, SiO_2 — 1622, $Al_2O_3 \cdot SiO_2$ — 1764, $MnO \cdot Al_2O_3$ — 1812 and Al_2O_3 — 1841.

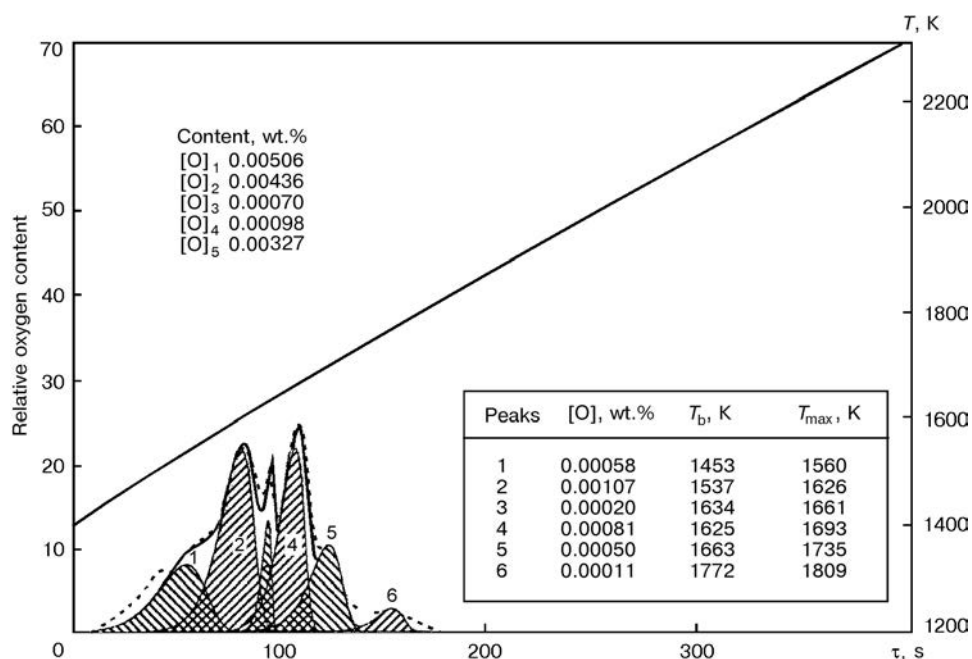
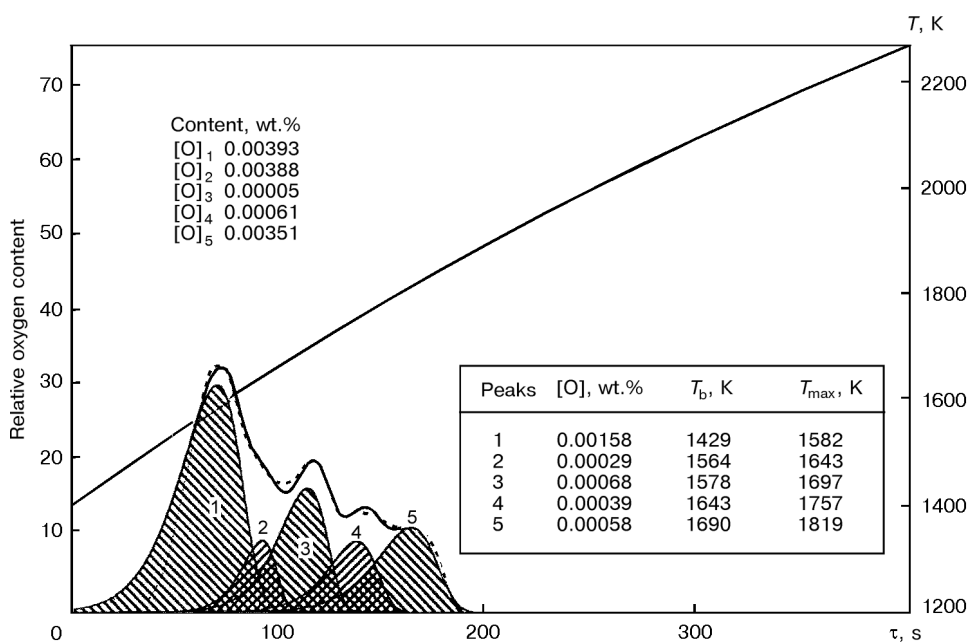
Sample 2-157 (Figure 3). $[O]_5 = 35.1 \cdot 10^{-4}$ wt.%. 5 peaks corresponding to the following types of inclusions were identified from the evalogram: $2FeO \cdot SiO_2$, MnO, $MnO \cdot SiO_2$, SiO_2 and $MnO \cdot Al_2O_3$.

Table 3. Mechanical properties of the commercially melted steel 110G13L castings

Heat No.	Tensile strength, kgf/mm ²	Yield stress, kgf/mm ²	Elongation, %	Reduction in area, %	Impact toughness KCU, kgf-m/cm ²
2-121	72.4	43.0	19.40	28.0	17.0
2-167	77.0	46.2	6.20	21.6	17.0
2-156	70.2	42.4	22.20	20.0	17.2
2-133	71.2	42.8	19.40	26.2	17.2
2-150	71.7	43.0	16.20	20.4	17.2
2-152	71.4	42.8	22.20	31.2	17.4
2-153	76.2	45.7	22.29	28.0	17.6
2-155	76.2	45.7	16.20	26.6	18.0
2-139	86.5	37.2	28.00	51.9	18.7
2-96	90.1	45.0	34.00	54.6	23.8

**Table 4.** Results of determination of the mass fraction of oxygen in steel 110G13L by the FGA method

Heat No.	Content, $1 \cdot 10^4$ wt. %				
	[O] ₁	[O] ₂	[O] ₃	[O] ₄	[O] ₅
2-167	86.3	70.0	16.3	6.9	66.6
2-156	74.2	56.1	18.0	13.9	43.4
2-133	325.5	325.5	—	10.5	297.1
2-150	50.6	43.6	7.0	9.8	32.7
2-152	126.6	110.0	16.1	7.9	100.5
2-153	161.5	121.5	40.0	10.2	108.9
2-155	532.3	510.2	22.1	5.5	500.2
2-157	39.3	38.3	0.5	6.1	35.1

**Figure 2.** Evalogram of fractional gas analysis of steel 110G13L (heat 2-150)**Figure 3.** Evalogram of fractional gas analysis of steel 110G13L (heat 2-157)

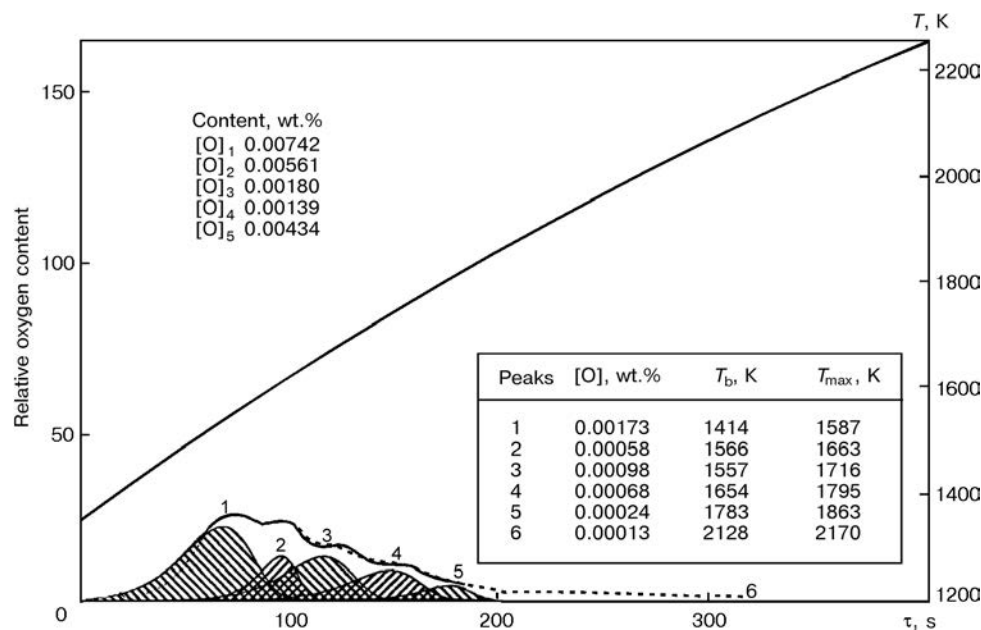


Figure 4. Evalogram of fractional gas analysis of steel 110G13L (heat 2-156)

Temperatures (T_b , K) were: $2\text{FeO}\cdot\text{SiO}_2$ — 1429, MnO — 1564, $\text{MnO}\cdot\text{SiO}_2$ — 1578, SiO_2 — 1643 and $\text{MnO}\cdot\text{Al}_2\text{O}_3$ — 1690.

Sample 2-156 (Figure 4). $[\text{O}]_5 = 43.4 \cdot 10^{-4}$ wt. %. The evalogram is represented by 6 peaks. The first peak corresponds to oxides FeO and SiO_2 ($2\text{FeO}\cdot\text{SiO}_2$), the second and third peaks — to MnO , the fourth and fifth peaks — to inclusions with a high alumina content, and the sixth peak — to corundum. Temperatures (T_b , K) were: $\text{FeO}\cdot\text{SiO}_2$ — 1349, MnO — 1592, SiO_2 — 1622, $\text{Al}_2\text{O}_3\cdot\text{SiO}_2$ — 1764, $\text{MnO}\cdot\text{Al}_2\text{O}_3$ — 1812 and Al_2O_3 — 1841.

Sample 2-167 (Figure 5). $[\text{O}]_5 = 66.6 \cdot 10^{-4}$ wt. %. The evalogram is represented by 8 peaks indicating to several oxygen compounds present in the sample: system $\text{FeO}\cdot\text{SiO}_2$, $\text{MnO}\cdot\text{SiO}_2$ and $\text{Al}_2\text{O}_3\cdot\text{SiO}_2$. Temperatures (T_b , K) were: $2\text{FeO}\cdot\text{SiO}_2$ — 1351, MnO —

1597, SiO_2 — 1626, Al_2O_3 — 1729, $\text{MnO}\cdot\text{Al}_2\text{O}_3$ — 1768, Al_2O_3 — 1780.

Sample 2-152 (Figure 6). $[\text{O}]_5 = 100.5 \cdot 10^{-4}$ wt. %. The evalogram comprises 15 peaks. Along with oxides similar to those identified in the previous samples, oxides of the $\text{Al}_2\text{O}_3\text{--}(\text{Ca}, \text{Mg})\text{O}$ system were revealed in sample 2-152. Temperatures (T_b , K) were: $2\text{FeO}\cdot\text{SiO}_2$ — 1334, SiO_2 — 1593, MnO — 1601, $\text{Al}_2\text{O}_3\cdot\text{SiO}_2$ — 1767, $\text{MnO}\cdot\text{Al}_2\text{O}_3$ — 1832, Al_2O_3 — 1864. Allowing for a complex spectrum of oxides in the inclusions and the presence of $(\text{Ca}, \text{Mg})\text{O}$, it can be assumed that some of them have an exogenic origin (slag inclusions).

Sample 2-133 (Figure 7). $[\text{O}]_5 = 297.1 \cdot 10^{-4}$ wt. %. The evalogram comprises 12 peaks. The first three peaks correspond to inclusions $\text{MnO}\cdot\text{FeO}$ and MnO , the next six peaks correspond to the $\text{MnO}\cdot\text{SiO}_2$ and

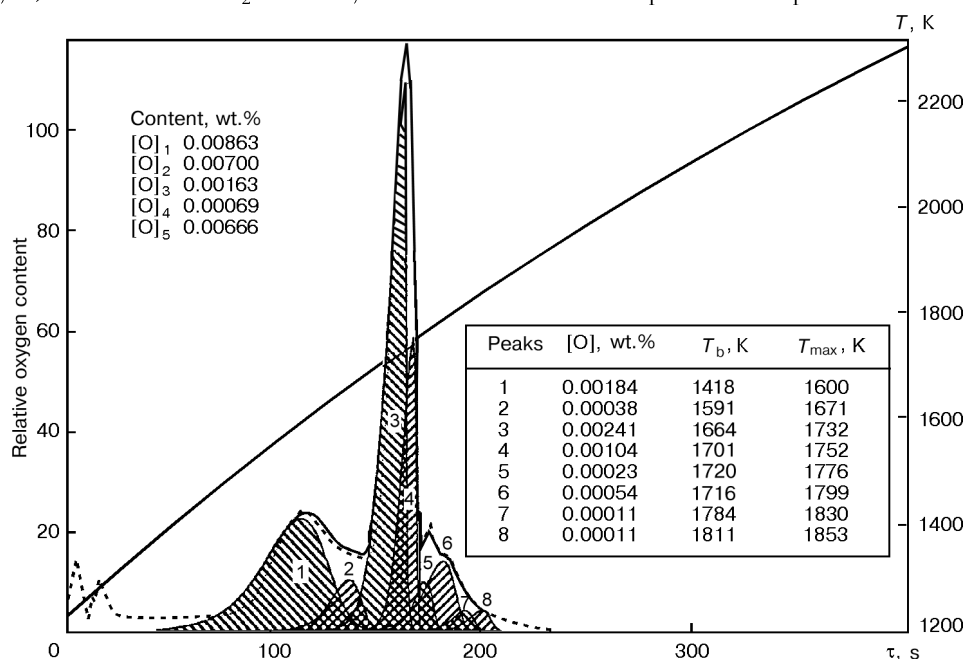


Figure 5. Evalogram of fractional gas analysis of steel 110G13L (heat 2-167)

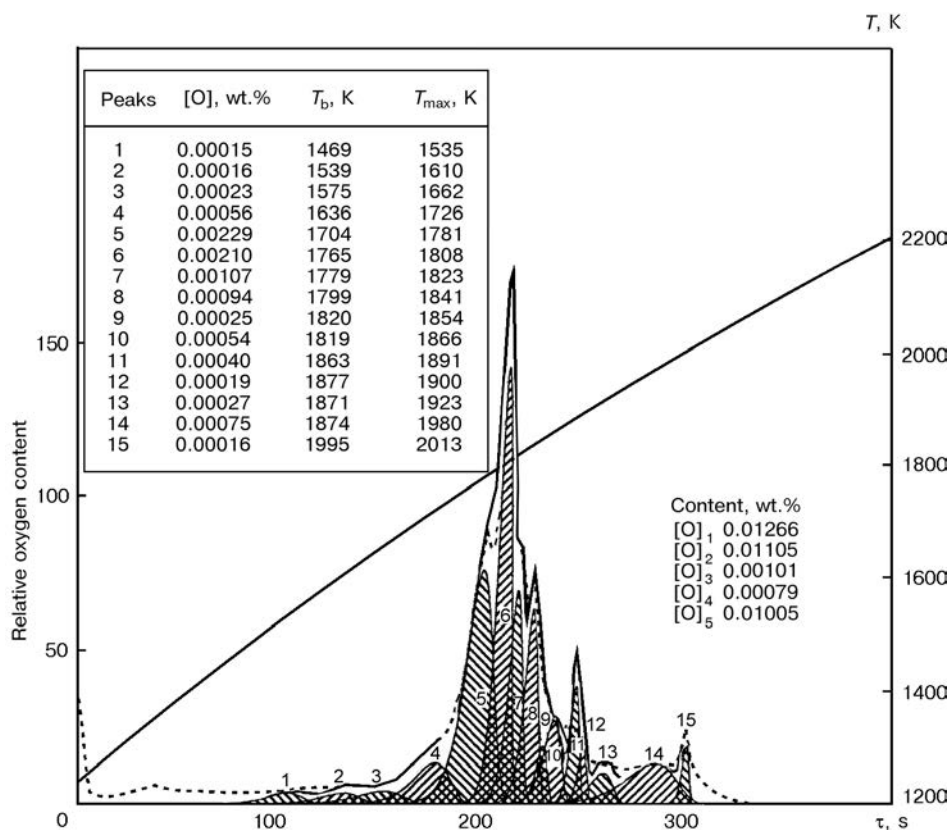


Figure 6. Evalogram of fractional gas analysis of steel 110G13L (heat 2-152)

SiO₂ system, and the last three peaks correspond to MnO·Al₂O₃ and Al₂O₃. Temperatures (T_b , K) were: 2FeO·SiO₂ – 1342, SiO₂ – 1611, MnO – 1611, Al₂O₃·SiO₂ – 1724, MnO·Al₂O₃ – 1771 and Al₂O₃ – 1781. Of note is the fact that reduction of all oxides is ended within a short period of time.

Sample 2-155 (Figure 8). [O]₅ = 500.2·10⁻⁴ wt. %. The evalogram comprises 12 peaks. The first four peaks are oxides FeO, MnO and SiO₂, the next six peaks are oxides MnO, SiO₂ and Al₂O₃, and two

latter peaks are Al₂O₃·SiO₂. Temperatures (T_b , K) were: 2FeO·SiO₂ – 1346, MnO – 1597, SiO₂ – 1615, Al₂O₃·SiO₂ – 1788, MnO·Al₂O₃ – 1847 and Al₂O₃ – 1886. The high total oxygen content of inclusions is indicative that part of them are of the exogenic origin, like in samples 2-152, 2-153 and 2-155.

It follows from the data shown in Figures 2–8 that, despite the steel melting technology accepted at the factory, keeping to the parameters of the oxi-

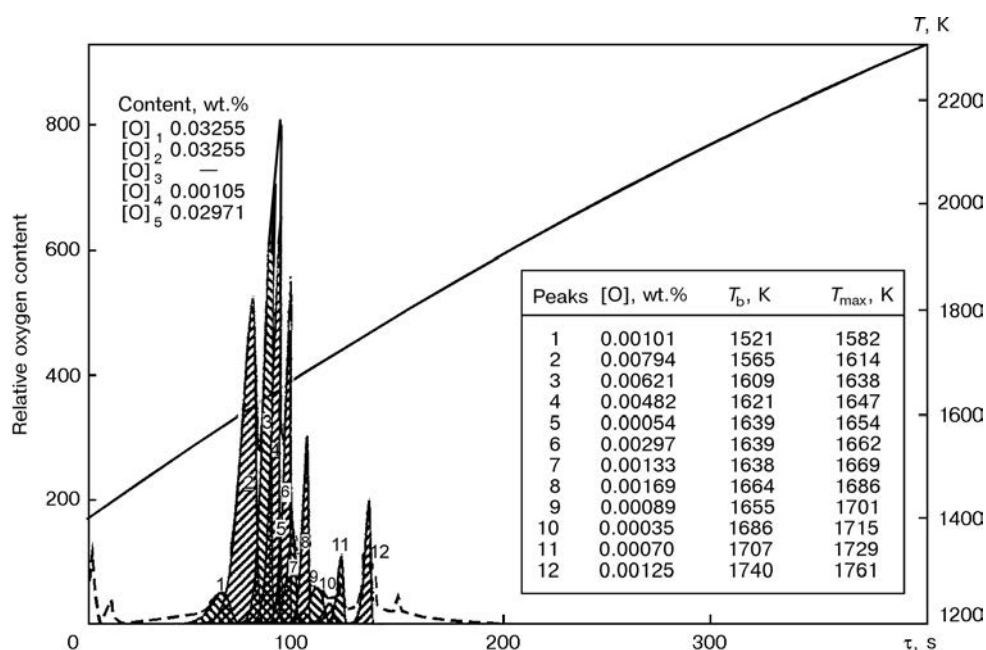


Figure 7. Evalogram of fractional gas analysis of steel 110G13L (heat 2-133)

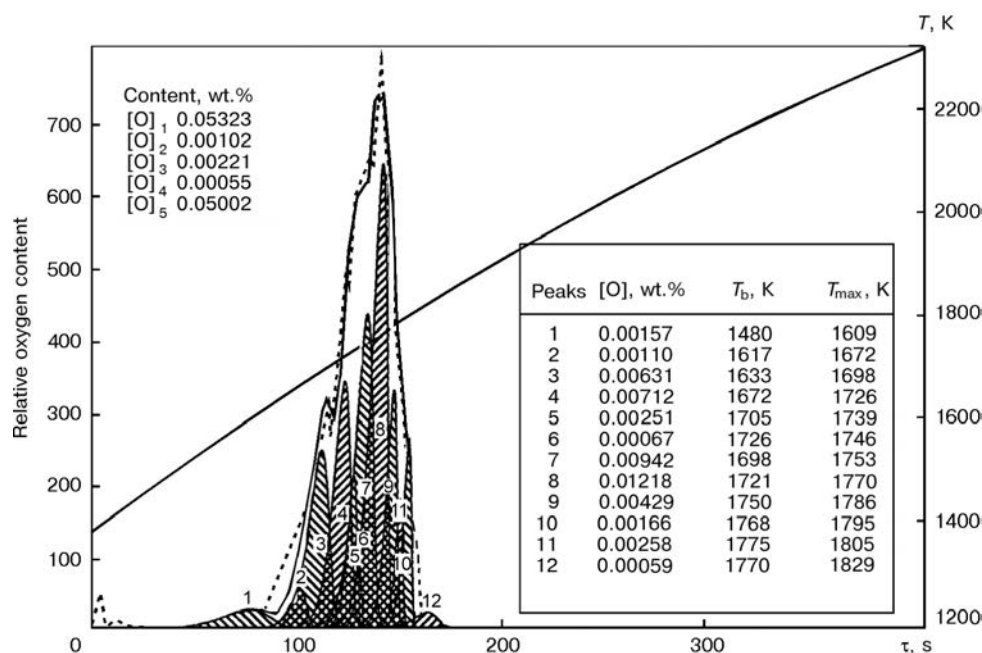


Figure 8. Evalogram of fractional gas analysis of steel 110G13L (heat 2-155)

dation period of melting, preliminary deoxidation of steel, alloying the steel pool with ferromanganese and final deoxidation of the steel in the furnace and ladle with aluminium, the amount of oxygen combined into non-metallic oxygen-containing inclusions (oxides, silicates, aluminates, etc.) in the investigated samples is characterised by a wide heat-to-heat variation. The steel samples 0.3–0.7 g in weight are naturally not entirely representative in terms of evaluation of the degree of contamination of the technology sample, i.e. bar, with non-metallic inclusions, all the more so the cast cores or frogs. Nevertheless, the results of evaluation of the oxygen content of inclusions characterise the entire spectrum of oxides present in the castings of this or that heat of the steel.

The samples of metal of heats 2-150 (Figure 2), 2-157 (Figure 3), 2-156 (Figure 4) and 2-167 (Figure 5), with the oxygen content of inclusions equal to $32.7 \cdot 10^{-4}$, $35.1 \cdot 10^{-4}$, $43.4 \cdot 10^{-4}$ and $66.6 \cdot 10^{-4}$ wt.%, respectively, contain inclusions which can be classed primarily with the endogenic type. In the rest of the steel samples (Figures 6–8) the high oxygen content and a wide spectrum of the gas evolution peaks are likely to be caused by ingress of slag inclusions during casting the metal into the moulds. Therefore, in addition to endogenic inclusions, the steel samples also contained a different amount of inclusions of the exogenic origin.

Comparative analysis of data on mechanical properties of metal of different heats and oxygen content of inclusions failed to reveal a definite relationship between these values. This proves the known postulate that mechanical properties of castings of the high-manganese austenitic steel 110G13L depend not only upon the melting process parameters, but also to a considerable degree upon the parameters of heat treatment of the castings.

Correction of parameters of melting and deoxidation of steel with aluminium was done on the basis of results of the study described in this article. Other parameters of the start-to-finish technology being kept to, this correction allowed the yield of the first group of the castings to be increased from 15–20 to 40–50 %.

In addition to the experiments on identification of the types of oxide inclusions, the TS-436 analyser was also used to evaluate the nitrogen content (Table 5).

Table 5. Nitrogen content of commercially melted steel 110G13L samples

Heat No.	Nitrogen content according to data of several evaluations, wt. %	Mean content, wt. %
2-96	0.0136; 0.0137; 0.0141	0.0138
2-121	0.0279; 0.0279	0.0279
2-133	0.0224; 0.0262; 0.0264; 0.0240; 0.0132	0.0224
2-139	0.0115; 0.0115	0.0115
2-150	0.0292; 0.0289; 0.0289; 0.0277; 0.0278	0.0285
2-152	0.0255; 0.0256; 0.0259	0.0257
2-153	0.0250; 0.0253; 0.0238	0.0247
2-155	0.0157; 0.0154; 0.0154; 0.0149; 0.0148	0.0152
2-156	0.0242; 0.0248; 0.0242; 0.0237; 0.0229	0.0239
2-157	0.0198; 0.0196; 0.0188; 0.0190	0.0193
2-67	0.0192; 0.0193; 0.0194	0.0193



As follows from the data of this Table, a minimum mean nitrogen content of the sample of heat 2-139 was 0.0115, and a maximum content was 0.0285 wt.% in the sample of heat 2-150. It is likely that a heat-to-heat variation in the amount of nitrogen should be taken into account also in analysis of complex relationships between mechanical properties of steel, melting technology and heat treatment of castings.

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