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ON NEW APROACH TO CONSTRUCTION OF THE ESR AUTOMATIC CONTROL SYSTEM

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On the basis of the melting nature analysis of consumable electrodes of small section under fluxes of different electric conductivity, the possibility is shown of «self-vibration» phenomena occurrence in electrodes in ESR.

Keywords: ESR, control system, hardening, consumable electrode end, slag resistance, vibration, mass-exchange processes

Electroslag remelting (ESR) is a well studied process for production of steels and alloys of the highest quality. ESR has been existing in present form for about 50 years, first of all due to works of scientists of the E.O. Paton Electric Welding Institute of NASU. However, this process is not completely investigated and understood either. For example, main physicalchemical reactions, which stipulate proceeding of the refining processes (removal of harmful impurities, gases and non-metal inclusions) proceed in the places of contact of the molten metal with the slag, i.e. on interface of two phases ---- the metal and the refining slag [1]. So, three stages (zones in which refining processes proceed) exist: on the consumable electrode end, in the zone of movement of the molten metal drops from the electrode through the molten slag to the metal pool, and on the boundary of the slag and the metal pools. At the same time data exist that indicate that main refining of the metal may occur not just on the electrode end, but also in the third stage (third zone of refining) ---- at of the metal--slag interface [2]. Intensity of refining depends upon many parameters: chemical composition and temperature of the slag, time of the molten metal reaction with the slag, rate of delivery of the reagents from the metal volume to the interphase surface and withdrawal of the reaction products into the slag volume, etc. [3--6].

State-of-the-art practice of ESR performance is targeted at application of its possibilities first of all for refining structure of the metal and producing ingots with minimal development of the segregation processes, because cleaning of the metal from gases, harmful impurities and non-metal inclusions can be rather efficiently performed in manufacturing of consumable electrodes for ESR. From this viewpoint the ESR process should be constructed in such way that remelting be performed under «quasi-stationary» conditions with minimally possible change of main parameters of the process, including lowering of the consumable electrode into the slag. Attempts of a number of national and foreign researchers to intensify refining processes by vibration of the consumable electrode, its rotation and other methods have to be considered at present from another viewpoint.

These methods bring rather good results only for ingots of small diameter on pilot and, in the best cases, pilot-commercial specimens. In case of transition to commercial ingots of steels and alloys, having propensity to segregation and big length of the biphase zone (for example, high-carbon tool steels and super-alloys), mentioned disturbances in the process of remelting inevitably cause intensification of segregation.

That's why state-of-the-art ESR control systems are arranged in such way that additional disturbances and violations, first of all in the hydrodynamics of the slag and the metal pools, in production of ESR ingots be excluded.

At the same time in all ESR control systems one of the main parameters, which is measured indirectly but used for the process control, is resistance of the slag pool. It is evident that this parameter varies in the process of remelting not just because of change of the slag chemical composition (which can be kept stable using different technological methods for a rather long time), but also because of constant movement of metal drops from the consumable electrode end to the interface of the slag and the metal pools. In this connection investigations, carried out in recent years in the field of ESR, in which slag containing metal calcium is used, allow finding new approach to construction of the ESR control systems [7--9].

These investigations are carried out for the purpose of increasing efficiency of ESR exactly as the process for cleaning the metal being remelted. ESR efficiency may be increased (from the viewpoint of refining of the metal from harmful impurities) by intensification of the mass-exchange processes between molten metal and slag by means of both increase of the molten film surface on the electrode end and reduction of the cycle of its renovation. This surface depends upon techno-

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logical conditions of melting, which determine shape and size of the lowered into the slag electrode end.

In particular, in performance of experiments with ESR having a standard control system, which always responses to change of the slag pool resistance by relocation of the consumable electrode, certain interesting, in opinion of the authors, facts were detected. For example, in remelting of electrodes of small diameters (up to 100 mm) using CaF_2 -Ca slag characterized by high electric conductivity [7-9], low deepening of the electrode was detected. It melted near the slag pool surface and its end remained practically flat.

The other remelting parameters being the same, lowering of the electrode into the slag increases as electric conductivity of the slag, used in the experiments, reduces in the following sequence: CaF₂--Ca, CaF₂, ANF-1P, ANF-6. The biggest lowering is characteristic of the ANF-6 flux. In this case the electrode has a cone end with a developed surface of the metal film, which allows (other conditions being the same) assuming possibility of deep refining. However, in practice it is fair only when comparison is made with remelting, in which ANF-1P and CaF₂ fluxes are used. The best refining was registered in remelting, using CaF₂--Ca flux [7, 10]. Probably, important part, together with other parameters, plays in refining processes also rate of the reaction surface renovation, i.e. rate of the film formation and break-off of the metal drops from the electrode end.

Thickness of the film and rate of the surface renovation may be effected by external mechanical methods, in particular rotation or vibration of the consumable electrode [11--13]. However, state-of-the-art practice of ESR does not allow using mentioned approaches because of their negative influence on metallurgical quality of the ESR ingots.

A number of experiments were carried out for determining possibilities of more accurate consideration of the influence of mentioned processes on ESR as an object of control.

Let us consider the mechanism of occurrence of vibrations, and the factors, which effect change of the electrode feed rate in ESR.

Periodic changes of the electrode holder movement speed (upper carriage of the ESR furnace) and, as a result, of the consumable electrode in ESR of ingots of small section are caused by fluctuations of the remelting current values due to periodic build-up of the drop and reduction of the slag pool resistance because of reduction in this place of the distance between the electrode end and the metal pool $h_{\rm sl}$. So, the metal drop, which builds up on the steel electrode end, reduces this interspace, causing short-time increase of the current. For example, in remelting of steel electrode of 70 mm diameter in slag pool from ANF-6 of 50 mm height excess of current at the instant of the drop break-off constitutes on average 20 % of the designed value of the remelting current. Standard system of the ESR automatic control causes reverse movement of the carriage, which feeds the consumable electrode, in order to bring current value of current to the preset one. It is connected with the fact that the system for controlling electrode feed rate has the remelting current feedback ---- $I_{rem} = I_{preset}$. The remelting current is determined by the fed voltage $U_{\rm rem}$ and the slag pool resistance R_0 ($I_{\rm rem} = U_{\rm rem} / R^0$). The remelting voltage has a relatively stable value, which practically does not change in the process of remelting (except switching over of the voltage stages and voltage fall at the electrode). Resistance of the slag pool can vary, because it depends upon both physical properties of the slag ($T_{\rm rem}$, chemical composition) and the $h_{\rm sl}$ value. It is accepted that at constant rate of melting $h_{\rm sl}$ remains invariable, and due to this stability of the remelting current is achieved. In reality it is fair only for averaged data.

Resistance of the slag is function of $h_{\rm sl}$ and specific resistance of the slag $R^0 \approx f(h_{\rm p}, \rho_{\rm sl})$. Then maximum change of the slag resistance during growing of the drop may be written as $\Delta R \approx f(h_{\rm dr}, \rho_{\rm sl})$, where $h_{\rm dr}$ is the drop length before its break-off from the electrode end. Level of the current mismatch will constitute

$$\Delta I = \frac{U_{\text{rem}}}{R^0} - \frac{U_{\text{rem}}}{R^0 - \Delta R} \Rightarrow \Delta I = \frac{U(R^0 - \Delta R) - UR^0}{R^0(R^0 - \Delta R)} \Rightarrow$$
$$\Rightarrow \Delta I = \frac{-U\Delta R}{R^0(R^0 - \Delta R)} \Rightarrow \Delta I = \frac{U_{\text{rem}}}{R^0} \frac{\Delta R}{\Delta R - R^0}.$$

Value $K = \Delta R / (\Delta R - R^0)$ directly determines degree of the mismatch. This coefficient, characterizing variation of the remelting current, determines



Figure 1. Substitution electrical scheme of ESR process: a --- in absence of drop; b --- with drop



Figure 2. Character of change of electrode movement speed in ESR using ANF-6 (*a*), ANF-1P (*b*), pure CaF_2 (*c*) and CaF_2 -Ca (8 %) (*d*) fluxes

change of h_{sl} as a result of the drop formation. Under «quasi-stationary» conditions it should be constant.

The current mismatch is connected directly proportionally with the rate of the consumable electrode feed rate. The higher is the current mismatch Δl , the higher is increment of the feed rate Δv .

It is difficult to draw conclusions, analyzing coefficient *K*, on the reasons of occurrence of the electrode vibrations in ESR, because ΔR is function of R^0 .

Let us present the ESR process in the form of the known substitution scheme with two resistances connected in parallel (Figure 1), whereby the left branch represents resistance of the slag pool at the initial instant (prior to formation of the drop), and the right branch corresponds to the instant of the drop formation and depends upon its geometric size $R_{\rm dr}$.

Then one may write

$$\frac{1}{R^0 - \Delta R} = \frac{1}{R^0} + \frac{1}{R_{\rm dr}}$$

Hence $\Delta R = R_0 / (R_{\rm dr} / R_0 + 1)$.

Let us assume that during melting of the electrode near surface of the slag pool (small deepening) resistance in the system before formation of the drop equals R'^0 , and in case of a greater deepening ---- R''^0 . As far as it is possible to assume that the forming drops have a certain similar size, then $R_{dr} = \text{const.}$ At the initial instant of melting, when the drop is absent, the electrode is in the medium position and $R'^0 > R^0 > R''^0$. As the discharge interval increases, R^0 increases too, tending to R'^0 . Then ΔR increases, which causes growth of |K|, whereby amplitude of the electrode movement gets bigger.

In case of h_{sl} reduction, R_0 decreases and tends to R''^0 . Then ΔR also reduces, and |K| and amplitude of the electrode movement get smaller.

So, possibility of occurrence of the conditions, under which the consumable electrode (carriage of the electrode holder) makes during its movement downwards reciprocation movements (vibrations), may depend upon deepening of the electrode into the slag.

Taking into account noted current fluctuations, disturbances caused by such fluctuations (vibration) of the electrode, may be rather noticeable and cause undesirable enhancement of segregation in the hardening ingot. Origination of mentioned phenomena may be controlled by constant monitoring of changes of the carriage movement speed in time.

For this purpose a special device was developed, which represented a unit consisting of two measurement systems, operating in parallel: an analogue voltage sensor and digital pulse sensor of angular rotation. Application of such combination is caused by peculiarities of possible technological conditions of remelting.

So, for example, it becomes difficult to use a digital sensor in case of periodic reversal movement of the electrode because of mismatch of the frequency of updating of information on speed and frequency of vibration of the electrode. At the same time application of the analogue sensor allows with sufficient degree of accuracy controlling reversal speeds of the



electrode movement and speed of its total movement. In case of absence of the vibration, application of the digital sensor allows presenting information in visual form.

Using the developed measurement system, change of the electrode movement speed in the ESR process under different fluxes (ANF-6, ANF-1P, pure CaF_2 and CaF_2 --Ca) was estimated, while all other technological parameters remained constant (current, remelting voltage, slag pool height, size of the electrodes, and coefficient of the mould filling).

It was established that character of the speed change, all other conditions being the same, is strictly connected with composition of the flux. So, in case of using ANF-6 flux, the electrode is fed into the slag uniformly, without noticeable fluctuations (Figure 2, a).

In remelting using ANF-1P flux and pure calcium fluoride pulsations of the electrode movement speed occur, achieving 30 (for ANF-1P) and 40 % (for CaF₂) of its average value (Figure 2, *b*, *c*).

Application of the metal calcium as additive to pure CaF_2 causes significant increase of pulsations of the electrode feed rate (Figure 2, d).

Electric conductivity of the ANF-6 flux is much lower than that of the flux containing metal calcium [8, 9]. That's why in case of its using for ESR the electrode is deepened into the slag at a rather big distance. In remelting, in which slag of the CaF_2 --Ca system is used, the electrode stays in surface layers of the slag pool, which causes its inevitable fluctuations in the course of melting ---- self-vibration.

At the same time state-of-the-art measurement instruments and high-speed computer facilities allow leveling considered phenomena and discarding them as undesirable «noise», which will not be accepted by the control system as a control signal and cause unnecessary movements of the consumable electrode in the process of remelting. In its turn it allows improving quality of the ESR metal.

The results obtained confirmed concept of a principally new ESR control system, designed in PWI in recent years, which is based on suppression of the consumable electrode self-vibration [14]. Application of this system on commercial ESR furnaces allows in remelting with any kind of flux rather precise fixing of minimal deepening of the consumable electrode into the slag. New control system successfully operates on the 2.5-ton ESR furnace and passes industrial check on the 10-ton furnaces.

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Specifications of the ASh 115 M machine



Thickness of weld edges, mm 40200
Quantity of electrodes, pcs 2
Diameter of electrode wire, mm 3.0, 4.0
Welding current per electrode at duty cycle = 100 %, A up to 800
Electrode feed speed, m/h 80-450
Electrode feeding scheme independent
Vertical displacement speed, m/h 0.6-12.0
Turning of electrodes in ESW of inclined joints, degree ±25
Correction of position of electrodes:
along the groove±60
across the groove ±20
Driving force of traverse gear, kg≥800
Oscillation amplitude, mm 100
Guide rail flexible strip with involute gear rack

CERTAIN ASPECTS OF DEPOSITED LAYER FORMATION IN PORTIONED ELECTROSLAG CLADDING WITH APPLICATION OF LIQUID ADDITIVE METAL

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Peculiarities of the deposited layer formation in electroslag cladding by liquid metal in the current-conducting mould are considered. Influence of the portioned pouring conditions on quality of the joint zone and crystalline structure of the deposited layer (formation under certain conditions of the undulated or rectilinear joint zone, conditions of constant increase of the penetration depth or general lack of penetration, formation of drops in the deposited layer, and changes of the crystalline structure caused by the portioned pouring of the liquid additive material) is shown.

Keywords: current-conducting mould, electroslag cladding, portioned pouring of molten metal, peculiarities of deposited layer formation

Electroslag process with separate melting and solidification of the metal [1--3], developed on the basis of application of the current-conducting mould as a nonconsumable electrode and a molten additive metal, allows overcoming rigid dependence between productivity of the classical electroslag process and its temperature parameters, and successful implementing of both cladding of different cylindrical items and melting solid and hollow ingots. In contrast to the canonical scheme of electroslag remelting or cladding, characteristic peculiarity of this process is at present stage of its development a portioned supply of the molten additive metal on the slag pool surface. Continuous supply of the molten additive metal is prevented by the difficulties connected with need to pour the molten metal at comparatively low rates (up to 1000 kg/h) and its thin (the diameter up to 5 mm) jet.

Approximate rates of steel casting in the form of a continuous jet of the molten metal through the holes of different diameter are as follows [4]:

Hole diameter, mm	2	3	4	5	6	7
Rate of casting, kg∕h	190	394	706	1087	1590	2226

Portioned supply of the molten additive metal stipulates a number of peculiarities of the technological process of melting of the ingots and cladding. Some of them, concerning the cladding, are described in [5]. It should be noted that exactly in cladding these peculiarities are especially numerous and can cause undesirable consequences. That's why one has to pay special attention to their study.

One of the main issues of formation of the metal deposited layer is character of the base metal penetration. In case of continuous supply of the molten metal into the cladding area and uniform movement

of the heat release source relative the surface being clad penetration of the base metal is mainly of rectilinear character detected, for example, in welds, produced by the electroslag welding, using a consumable electrode under stationary conditions of melting. In this case supply of the metal with the drops into the pool is of discreet character too, but high frequency of fall of the drops and their small mass allow considering this flow of the molten metal as a continuous jet. In portioned supply of the molten additive metal, when mass of a portion is measured by kilograms and frequency of feeding of the portions equals 1 min^{-1} , another problem occurs. At constant linear rate of cladding a stepwise movement of the heat release source over the height of the surface being clad occurs at the frequency of the metal portion feeding. It can not but effect character of the penetration.

In addition, interrupted feeding of the molten metal to the solidification front at constant rate of solidification stipulates stepwise change of the main solidification parameters, which inevitably causes periodic change of the crystalline structure.

Usually compositions of the cladding and the base metal differ from each other. Constant character of the metal pool hardening and discrete character of inflow into it of the metal of different composition inevitably causes repeated inhomogeneity of chemical composition of the metal over height of deposit.

One more peculiarity of the deposited layer formation at portioned feeding of the additive metal is manifested in the case, when melting point of the metal being clad is significantly lower than melting point of the base metal. Then within the time span between pouring of the additive metal portions the metal is fed into the pool, which drains from the melted surface of the base metal and has a higher melting point than temperature of molten metal of the metal pool. In certain situations this can cause formation of so called drops, i.e. particles of a more refractory metal, which did not have time to dissolve and solidify in the metal pool [6].

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Figure 1. Fragment of joint zone — profile of penetration (a) — and melting of axle by slag (b, c) at portioned feeding of liquid additive metal in process of electroslag cladding (ESC LM): 1 — area of metal pool contact with billet being clad; 2 — melting of axle in slag pool; 3 — metal pool

This work is devoted to investigation of these peculiarities.

Despite accumulated experience and volume of the carried out investigations, mechanism of the joint zone formation in portioned electroslag cladding by molten metal in the current-conducting mould is not completely clear. Available information is of ambiguous and sometimes of controversial character. In particular, results of physical [7] and mathematical [8] modeling of distribution in the slag pool of current, voltage, and heat capacity contradict in certain aspects each other.

Full-scale experiments on cladding of the rolls and investigation of the metallographic sections and sulfuric imprints of the longitudinal templates, taken from these rolls, allow drawing the following conclusions:

• profile of penetration is in majority of cases undulated (Figure 1, *a*);

• identity of the penetration profile before and after pouring a portion of the molten additive metal (Figure 1, b) prove that it is determined by character of the axle melting in the slag pool. Influence of the molten metal feeding is insignificant;

• at the deposit thickness 25--85 mm and diameter of the rolls up to 740 mm beginning of penetration of the axle and its maximal depth in the process of cladding are located in the place of the metal pool contact with surface of the billet (Figure 1, *c*);

• height of the penetration (melting of the axle over its height) is significantly smaller than depth of the slag pool;

• depth of the slag pool practically does not effect the form of penetration.

In order to explain mentioned phenomena let us consider hypothetical distribution of current in the slag pool during cladding of the axle in the currentconducting mould. In electroslag cladding only a portion of the slag pool is located in the current-conducting section of the mould. Its other part is located in the separation and forming sections of the mould. Current runs from the current-conducting section wall in two directions: horizontal component is directed to the axle being clad, while vertical component ---- to the metal pool. Absolute value of these components is determined by electric resistance of the slag pool to passage of the current in either direction. Due to the heat release during passage of the current through slag, the slag pool is maintained in molten state. However, an electric circuit may exist, which, evidently, exerts significant influence on the penetration and passes from the current-conducting section to the axle being clad through walls of all sections of the mould and zone of separation of the slag and the metal pools, which practically do not render resistance to passage of the current in contrast to the places of transition through the slag from one section to the other (Figure 2). As show the results of physical and mathematical modeling [7, 8], exactly in these places local areas of significant heat release were detected. Absolute value of this part of vertical component of the current may be comparatively small, but concentrated in small area of the axle surface adjacent to the metal pool (or to the metal inoculation at the beginning of



Figure 2. Areas of the highest heat release in slag pool in ESC in current-conducting mould: *1* — axle; *2* — current-conducting section; *3* — slag pool

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Figure 3. Hypothetic form of axle melting in slag at big (1) and small (2) gaps between axle being clad and mould (a), and scheme of axle penetration increase (b)

the process). That's why current density in this place may significantly exceed density of horizontal component of the current of the rest portion of the axle dipped into the slag. Due to melting the axle acquires the form, schematically shown in Figure 3, a. Resistance of this circuit does not depend upon the slag pool depth, but it is determined only by the level of transition resistances between the mould sections, which explains registered in the experiments independence of the axle melting character upon the slag pool depth at the invariable current. One has, evidently, to expect change of the axle melting character in case of significant reduction of the gap between the mould and the axle. Then, by means of reduction of the slag resistance in this area and increase of the horizontal component, role of the latter in melting of the axle will significantly increase.

So, in contrast to the classical process of electroslag welding, distribution of current in the slag pool of the current-conducting mould enables formation on the slag--metal boundary of the concentrated source of heating, whereby a certain role in formation of the fusion line play size and frequency of the molten additive metal feeding. While electrical conditions and rate of cladding determine average depth of pene-



Figure 4. White strips in macrostructure of deposited metal

tration, frequency and size of the portions determine amplitude of the penetration depth fluctuations. At invariable speed of cladding size of the portions and frequency of their feeding are rigidly connected. The bigger are the portions, the lower is frequency of their feeding and the bigger is the time interval between inflows of the additive metal into the metal pool. As a result local depth of penetration increases, and subsequent big portion of the metal drastically brings level of the metal pool into the position, where penetration is minimal, and the cycle repeats again. So, the smaller is the portion and the higher is the frequency, the lower is the amplitude of the penetration depth fluctuations. It is also evident that absolute value of the amplitude of the penetration fluctuations reduces as average depth of the penetration decreases. In case, when average depth of penetration is minimal and frequency of feeding of the portions is sufficiently high, undulated penetration may transit into the rectilinear one. It should be noted that at a certain ratio of average penetration depths, frequency and size of the portions, rectilinear penetration is also possible at sufficiently big average depth of penetration. One can see from Figure 3 that when the portion equals h_1 , undulated penetration will be formed. As size of the portions reduces and their frequency remains the same, i.e. as cladding productivity reduces, height of the wave crest will reduce, and absolute depth of penetration δ will increase, achieving its maximal level within the period between the portions in the point A, which corresponds to $H_{por} = h_0$ (Figure 3, b). If size of the portions is reduced, for example, down to h_0 and simultaneously frequency of their pouring is increased and the same productivity is preserved, increase of the average penetration depth will not occur and point A will move upwards over the undulated line. Profile of such undulated line will depend upon size of the portion (the lower is the portion, the more this profile will correspond to the

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Figure 5. Etching for dendrite microstructure of white strip zone and adjacent to it areas

straight line). Taking into account sluggishness of the electroslag process, even at the portioned feeding of the additive metal one can get straight fusion line. At the portion size $h_2 > h_1$ and preservation of the pouring frequency, i.e. in case of increasing the productivity formation of lacks of penetration is possible.

It follows from mentioned above that by change of size of the portions and frequency of their feeding in strict correspondence with average depth of penetration, determined by energetic parameters of the melting, it is possible to control shape of the fusion line in portioned electroslag cladding in the currentconducting mould, getting both a rectilinear fusion line and an undulated one with different heights of the wave.

However, even production of the rectilinear fusion line does not remove stepwise change of the solidification parameters at the instant of the metal portion pouring. This may cause formation of structural inhomogeneity in the deposited metal, manifested in formation of white strips in its macrostructure (Figure 4) and the most clearly seen in the peripheral areas. It is characteristic that the strips are not detected on the microsections, and microstructures of the metal areas within and outside white strips are identical.

One should assume that local changes of thermal conditions cause just local changes in orientation of growth of the crystallites (Figure 5), which causes change in these places of the light reflection intensity. Hardness of the areas is the same: *HRC* 42, 44, 42, 45.5 in the area of former white strips, and *HRC* 41, 45.5, 43 in the area outside white strips. That's why these inhomogeneities can hardly be used as a rejection sign. At the same time this issue requires for additional investigations.

More serious is the problem of chemical inhomogeneity of the deposited metal because of discrete character of the additive metal feeding to the continuously moving solidification front. The problem can be also aggravated by the fact that within the periods between feeding of the additive metal portions, the metal of different composition continuous to enter into the metal pool due to melting of the base metal.





Figure 6. «Drops» of base metal in deposited layer

In meantime in investigations of chemical composition and microstructure of the high-speed steel deposited on the axle from the low-alloy steel at penetration up to 10 mm, significant differences between different deposit zones were not detected. This fact can be explained by averaging of the metal pool composition due to its intensive mixing during rotation. However, there are cases when not the whole metal, which drains from the melted surface into the metal pool, manages to completely dissolve in it. It forms metal inclusions of the base metal in the matrix from the deposited metal (Figure 6). Confirmation of this are data of spectral analysis of the drop area for presence of vanadium (3.2 %), which give intermediate value between content of this element in the base (in 40KhM is not available) and additive (high-speed steel ---- 7 %) metals. Role of these inclusions in serviceability of the items is not quite clear.

One could draw aprioristic conclusion that such inclusions can reduce wear resistance of the items, for example, rolls of the rolling mills with a deposited layer from a high-speed steel, because these inclusions are softer, and exert positive influence on resistance to brittle fracture, inhibiting development of the cracks. However, industrial tests of the high-speed rolls do not confirm this. This issue requires for further investigations. Presented data on formation of the deposited layer in portioned electroslag cladding should be taken into account when developing melting conditions.

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MODELING OF EVAPORATION PROCESSES OF ALLOYING ELEMENTS FROM Ti--6Al--4V ALLOY IN EBCHM AND PORTIONED DISCHARGE

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Mathematical model of evaporation of alloying elements in electron beam cold hearth melting (EBCHM) and portioned discharge of the metal for Ti--6Al--4V titanium alloy is developed. Within the framework of the suggested model dependencies of chemical composition of the ingot being melted upon technological parameters of the process are obtained. Adequacy of the model is confirmed by the experimental data obtained as a result of melting of the Ti--6Al--4V alloy ingots.

Keywords: electron beam melting, intermediate unit, portioned discharge, chemical composition, evaporation

At present technological scheme of electron beam melting (EBM) with intermediate unit (EBCHM), which allows complete separating of the melting and refining processes, proceeding in the intermediate unit, and hardening of the metal in the mould, is widely used [1, 2].

To improve structure of the ingot a technology of electron beam melting with portioned discharge of the molten meal from the intermediate unit into the mould was developed in the E.O Paton EWI [3, 4], which has a number of essential advantages in comparison with traditional EBM technologies. Small vol-



Figure 1. Scheme of electron beam melting with intermediate unit (for designations see the text)

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ume of the molten metal portion, fed from the intermediate unit into the mould, suppresses convection processes in the melt during its solidification, which prevents development in the ingot of zonal segregation and formation of defects of the shrinkage origin.

As far as the melting process proceeds in vacuum, different pressure of alloying component vapor of the alloys can cause change of chemical composition of the ingot. So, an important aspect of the EBCHM process optimization is ensuring of correspondence of the ingot composition to the established standards.

Mentioned task of optimization was solved by the mathematical modeling method on the basis of the EBCHM process of the Ti--6Al--4V titanium alloy ingots. Suggested model allows getting dependences of chemical composition of the ingot upon different technological parameters of the EBCHM process.

EBCHM technology with portioned discharge. EBCHM scheme is presented in Figure 1. The ingot being melted 1 is fed at a constant speed into the zone of action of electron beam guns, where its melting occurs. Molten metal from the billet end being melted is fed into the intermediate unit (IU) 2, where its homogenization and refining from non-metal inclusions takes place. By means of the IU filling with the melt, the metal, solidified in the spout 3, is melted and a portion of the molten metal is discharged into the mould 4. By means of build-up of the ingot the mould pallet is lowered downwards. Within the whole period of melting by the electron beam heating a pool of molten metal is maintained in IU and the mould.

Description of processes of the melt component evaporation into vacuum. Evaporation of the alloying elements from the melt surface occurs in three main stages: mass transfer from the metal volume to the interface surface; change of the aggregate state on the interface surface; mass transfer of the element from the interface surface into the gas phase volume [4].

Total flow of the *i*-th element of the alloy (aluminium, vanadium, titanium), evaporating from the interphase surface, is calculated from the following relation [5]:



$$J_i = \beta_i SC_i,$$

where $\beta_i = \left(\frac{1}{\rho\beta_m^i} + \frac{1}{k_i^e}\right)^{-1}$ is the total coefficient of mass transfer, kg/(c·m²); ρ is the melt density, kg/m³; β_m^i is the mass transfer coefficient of the melt element m/s; k_i^e is the evaporation rate constant of the melt element, kg/(c·m²); *S* is the interphase surface area, m²; C_i is the weight share of the element in the molten metal volume.

Constant of the evaporation rate k_i^e is calculated according to the Langmuir equation [5, 6]:

$$k_i^e = \alpha_i \gamma_i p_i^0 \frac{M_{\mathrm{Ti}}}{\sqrt{2\pi RTM_i}},$$

where M_{Ti} is the molar mass of titanium (the alloy base element), kg/mol; M_i is the molar mass of a respective alloying element, kg/mol; α_i is the condensation coefficient; γ_i is the activity coefficient of the *i*-th element in the melt; p_i^0 is the equilibrium pressure of the element vapors above the molten metal surface, Pa; R = 8.314 J/ (kg·K) is the universal gas constant; *T* is the absolute melt temperature, K.

Dependences of the activity coefficient upon temperature and composition of the investigated system alloy can be calculated within the framework of the Redlich--Kister model [7].

Temperature dependences of the equilibrium pressure of vapors of the Ti--Al--V triple system elements are expressed by the following relations [8]:

$$\log (p_{\text{Ti}}^0(T)) = -\frac{22946}{T} + 10.581 - 0.373 \log (T);$$

$$\log (p_{\rm Al}^0 (T)) = -\frac{16379}{T} + 9.979 - 0.335 \log (T);$$

$$\log (p_{\rm V}^0(T)) = -\frac{24378}{T} + 10.762 - 0.325 \log (T).$$

As far as pressure of the vanadium vapor is much lower than that of aluminium (at the temperature about 2000 K by several orders) and, respectively, gradient of this element concentration in surface layer of the melt is insignificant, one may consider that vanadium is uniformly distributed over the whole volume of the melt.

Coefficient of the aluminium mass transfer in molten titanium equals $\approx 1.10^{-5} \text{ m/s} [5]$.

For titanium as main element of the melt coefficient β_m^{Ti} may be considered infinitely big.

Modeling of processes of alloying element evaporation from the billet end. In the process of the charge melting in IU a film of molten metal is formed on the billet end, from the surface of which evaporation into vacuum occurs. Melting plane is at the angle about 90°, that's why the melt is held on it mainly by the surface tension forces. It means a small overheating of the metal, because due to the temperature increase coefficient of the surface tension reduces and molten metal drains off into IU. Because of this reason temperature of melt in the film on the billet end being melted may be considered constant and approximately equal to the melt melting point.

In order to determine chemical composition of the metal, which gets into IU from the molten billet end, it is necessary to analyze mass exchange in the molten metal area. So, into the pool on the end the metal of known composition arrives at a known mass rate w_0 , weight share of the elements (aluminium, vanadium, titanium) [Me]₀ in which is determined by composition of the billet being remelted. From the molten metal surface evaporation into vacuum occurs, and into IU drains off metal of unknown composition [Me]_t.

Assuming that concentration of the alloy elements in the main part of the pool metal on the end (except a thin near-surface layer) is the same and equal to the final one $[Me]_t$, change of each element mass m_i in the pool within time *t* is calculated as follows:

$$\frac{dm_i^t}{dt} = w_0[\mathrm{Me}]_0 - \beta_i S_t[\mathrm{Me}]_t - w_t[\mathrm{Me}]_t, \qquad (1)$$

where S_t is the area of the molten metal surface on the end of the billet being melted, m²; w_t is the mean mass rate of the molten metal arrival into IU, kg/s.

The main share of the time this process proceeds in stationary mode. So, derivative of mass in time in the expression (1) identically equals zero:

$$w_0[Me]_0 - \beta_i S_t[Me]_t - w_t[Me]_t = 0.$$
 (2)

Taking into account that $[Al]_t + [V]_t + [Ti]_t = 1$, unknown $[Me]_t$ and w_t are found from the formula (2).

As an example of solution of similar system of equations dependence of weight shares of aluminium and vanadium upon mass rate of the charge feeding is presented in Figure 2. These dependences show that as mass rate of the ingot feeding increases, weight share of aluminium in the metal, which gets into IU, smoothly increases too, while content of vanadium in the melt reduces.

Modeling of evaporation processes during seasoning of the melt in IU. By means of the IU filling with the metal draining from the molten end of the billet, from the melt surface in IU heated by the electron beams, evaporation of the alloying elements into vacuum also occurs. In contrast to the molten metal pool on the billet end, continuous outflow of the metal does not occur from IU, that's why equations of the material balance for the melt elements in IU have the following form:

$$\frac{dm_i^e}{dt} = w_t [\text{Me}]_t - \beta_i S_e [\text{Me}]_e, \qquad (3)$$

where S_e is the area of melt in IU, m^2 ; m_i^e is the mass of the *i*-th element, which is in IU at the present





Figure 2. Dependence of weight share of aluminium (*a*) and vanadium (*b*) in metal draining from melted end of billet upon mass rate of charge feeding w_0 , [Al]₀ = 7.4 %; [V]₀ = 4.0 %

instant of time, kg; $[Me]_e$ is the weight share of a respective element in the IU melt.

Time of seasoning, i.e. the time between two consecutive discharges of the metal into the mould depends, in its turn, upon rate of the charge feeding (and, respectively, rate of the metal feeding into IU) and amount of the portion being discharged. Duration of this interval is approximately calculated according to the following expression:

$$t_k = \frac{m_p}{W_t},$$

where m_p is the mass of the portion discharged from IU, kg.

Before beginning of melting the skull of known chemical composition is arranged in IU, in which a small pool is induced. Later the metal from the molten billet end gets into this melt. In addition, after discharge of the portion a certain part of molten metal, which is mixed with molten metal of the billet, remains in IU. Proceeding from this fact the initial conditions, necessary for solution of the differential equation (3), are built.

Respectively, initial conditions for expression (3) at first discharge from IU will be as follows:

$$m_i^e\Big|_{t=0} = m_R[\mathrm{Me}]_{g}, \tag{4}$$

where $[Me]_g$ is the weight share of the alloy elements in the skull; m_R is the mass of molten metal in IU before arrival of the metal from the molten end of the billet, kg.

Solution of the equation (3) system, taking into initial conditions (4) and the fact that $[Al]_e + [V]_e + [Ti]_e = 1$, allow getting mass of molten aluminium, vanadium and titanium in the IU pool at the instant of discharge, i.e. at the instant of time t_k . Proceeding from these results weight share of these elements in the discharged portion is calculated.

Mathematical description of the following discharges from IU differs only by initial conditions. So, in expression (4) instead of weight shares of elements in the initial skull weight shares of the elements by the time of the last discharge from IU will be used.

Dependences of weights shares of aluminium and vanadium in the melt, contained in IU, upon ordinal number of the discharge are presented in Figure 3. It is characteristic for these dependences that after several discharges the content of these elements remains constant, and the process proceeds under stationary conditions.

Of a certain interest are dependences of stationary content of the alloying elements in the melt, contained in IU, upon time of seasoning. It should be noted that this time is not a controlled technological parameter, but the function of the discharged portion mass and mass rate of the billet feeding. The calculations showed that change of chemical composition of the discharged metal in case of variation of the portion mass is insignificant. At the same time change of the



Figure 3. Dependencies of weight shares of aluminium (*a*) and vanadium (*b*) in skull of IU upon ordinal number of discharge *N*; initial content of alloying elements in skull ---- [AI] = 6 %; [V] = 4 %

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Figure 4. Dependencies of weight shares of aluminium (a) and vanadium (b) in skull of IU upon time of seasoning

rate of the charge feeding gives practically linear dependences of stationary content of the alloying elements in the IU melt upon time of seasoning (Figure 4).

Modeling of processes of evaporation from pool surface of the molten metal in the mould. In contrast to IU, metal enters into the mould only at the instant of the portion discharge from IU, at once after which the following processes occur in the mould pool: mixing of the discharged portion melt and the metal contained in the mould; hardening of a portion of the metal; evaporation from the pool surface of the molten metal. The slowest from three listed processes is evaporation of the melt elements into vacuum, the quickest ---- mixing of the metal. And in comparison with characteristic time of evaporation (1--2 min) one may neglect duration of solidification and mixing.

That's why system of the material balance equations for this stage of EBCHM will look as follows:

$$\frac{dm_i^c}{dt} = -\beta_i S_c [\text{Me}]_c, \qquad (5)$$

where m_i^c is the mass of the *i*-th element located in the mould at present instant of time, kg; S_c is the mould area, m^2 ; [Me]_c is the weight share of a respective element in the mould melt at present instant of time.

Initial conditions for system (3) are created as follows. After first discharge a portion of the metal,

having got on the pallet being cooled, solidifies at once. The rest part is maintained in molten state by the system of electron beam guns, which heat the mould. Respectively, at the initial instant of time mass of the melt elements is determined by their weight share in the portion being discharged $[Me]_e^1$ and general mass of molten metal in the mould m_c . So, initial conditions for system (3) for first discharge from IU into the mould acquire the following form:

$$m_i^e\Big|_{t=0} = m_c [Me]_e^1.$$
 (6)

So, solution of the system of differential equations (5) with initial conditions (6), taking into account the fact that $[AI]_c + [V]_c + [Ti]_c = 1$, allows getting weight share of the elements in the mould pool by the time of second discharge, i.e. in time interval t_k .

Initial conditions for subsequent discharges are obtained as follows. After discharge of the *j*-th portion of metal from IU with weight share of respective elements in it $[Me]_{e}^{j}$, a part of molten metal in the mold hardens, and in molten state remains metal of mass m_c , which is a function of the technological melting parameters. Initial conditions for the time interval between discharges of the *j*-th and (j + 1)-th portions has the following form:



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Figure 5. Dependence of weight shares of aluminium (a) and vanadium (b) upon height of ingot h

(7)



Figure 6. Dependencies of stationary weight shares of aluminium (a) and vanadium (b) upon mass rate w_0 of charge feeding

So, system (5) with initial conditions (6) and (7) allows calculating content of the alloy elements in the solidified ingot metal over its whole height. These dependences for content of aluminium and vanadium are shown in Figure 5. Behavior of the presented dependences is similar to achievement of stationary conditions of chemical composition in IU, but stationary state is achieved in the ingot somewhat later (at 3--5 discharges of metal from IU).

In Figure 6 dependences of stationary values of weight shares of aluminium and vanadium in the ingot upon mass rates of the billet feeding are shown. Behavior of these curves is significantly different. While value of stationary weight share of aluminium increases as rate of the billet feeding gets higher, content of vanadium, on the contrary, reduces.

Check of adequacy of the proposed mathematical model. For checking obtained within the framework of the presented mathematical model results a series of experimental melts of cylindrical ingots from Ti--6Al--4V titanium alloy of 200 mm diameter was performed at the following technological parameters [5]:

power of electron beam guns which heat IU --- 290 kW;

• power of electron beam guns which heat the mould ---- 24 kW;

- IU width ---- 660 mm;
- IU length ---- 330 mm;

• weight share of aluminium in the billet being melted ---- 7.41 %;

 \bullet weight share of vanadium in the billet being melted ---- 3.94 %.

Comparison of the chemical analysis data of the samples, taken from the melted ingots, with results

of the calculation (Figure 6) showed sufficiently high accuracy of the suggested model.

CONCLUSIONS

1. Mathematical model, which allows describing process of evaporation of the alloying elements in electron beam melting of the Ti--6Al--4V titanium alloy ingots, is developed.

2. Within the framework of the suggested model character of influence of such technological parameters of the melting process as rate of the charge feeding, number of the metal discharges into the mould, and time of the melt seasoning in IU on chemical composition of the metal at each stage of the melting is investigated.

3. Experimental data, obtained in a series of melts, showed high accuracy of the developed model.

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HARD EROSION-RESISTANT GRADED COATINGS DEPOSITED IN VACUUM

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Possibilities of using electron beam technology and equipment for application of hard erosion-resistant graded coatings on components of state-of-the-art machinery are shown. Results of investigations of chemical composition, structure, and functional properties of hard erosion-resistant graded TiN- TiC-, and B_4C -base coatings deposited from vapor phase on surface of steel and titanium alloys according to the single-stage technology are presented.

Keywords: electron beam evaporation and deposition, hard erosion-resistant graded coatings, gas turbine blades

State-of-the-art technologies of high-rate electron beam evaporation and condensation in vacuum are widely used for deposition of the metal heat- and corrosion-resistant and thermal barrier coatings with an outer ceramic layer on components of the gas turbine engines [1--3]. At the same time the technologies and electron beam equipment for application of the functionally graded, including hard erosion- and wear-resistant coatings, were developed [4, 5]. Results of the investigations, carried out in the E.O. Paton Electric Welding Institute International Center for Electron Beam Technologies directed at development and perfection of the technological processes, designed for application of hard erosion-resistant graded coatings on the protected substrates, were presented. Interest to these coatings is stipulated by the possibility of their use as both erosion-resistant coatings and an outer erosion-resistant layer in the damping coatings on the gas turbine engine components [4, 6, 7].

Compressor blades (from the titanium- and ironbase alloys) of state-of-the-art turbines operate under conditions of action of a number of external aggressive factors (cyclic mechanical, gas-abrasive, and corrosive ones). As wear- and erosion-resistant coatings for their protection the coatings on the basis of nitride compounds, applied, as a rule, by the method of plasmachemical synthesis with application of the vacuum-arc technology, are used [8--10]. The electron beam technology of the hard erosion-resistant coating deposition is attractive from the viewpoint of both high rates of the coating deposition (up to 1 μ m/s) and improvement of the structure and properties of the coatings being synthesized.

Direct electron beam heating and evaporation of the refractory compounds (oxides, carbides, borides, and nitrides) is accompanied by their partial dissociation. In condensation of the vapor flow on heated up to the necessary temperature substrate reverse process of the recombination and recovery of the crystalline structure and stoichiometry takes place, provided chemical composition of the vapor flow remains identical to the initial substance composition.

A number of compounds, for example, WC, TiN, ZrN, can not be deposited by means of direct evaporation, because during heating they decompose with formation of the products with drastically different volatility. Mentioned compounds in the form of condensates can be produced as a result of evaporation of the components from two independent sources or so called reactive evaporation with introduction into the working chamber of a respective gas and ionization of the vapor flow [11], whereby of great significance is ensuring of high level of the coating adhesion to the steel or titanium substrate, the heating temperature of which should not exceed 500 °C for preservation of respective physical-chemical properties of the substrate material.

In Figure 1 scheme of the ion-activated electron beam evaporation and condensation in vacuum, used for production of the erosion-resistant coatings, is presented. Simultaneously with evaporation by the electron beam of substances from the molten pool by means of the annular electrode the arc discharge is excited (voltage about 50 V, current up to 100 A) in the vapor flow between the electrode and the pool surface, and to the substrate negative electric potential (bias) up to 500 V is fed. Tending to the substrate positive ions, formed in the partially ionized vapor flow of the substance being evaporated, acquire an additional energy and increase energy of the pulse during collision (adsorption) with the substrate.

For production of a number of functionallygraded, for example, titanium nitride base coatings, and synthesis of the coatings a reactive gas (nitrogen) is introduced into the ionized vapor flow. An auxiliary water-cooled crucible, located adjacent to the main crucible, ensures, if necessary, possibility of evaporation of some additive by the independent electron beam and its introduction into the main vapor flow.

Rectangular specimens of $22 \times 8 \times 2.4$ and $20 \times 15 \times 4$ mm size, made from the VT1-0 titanium alloy and steel of the following chemical composition, wt.%: 24 Ni, 14.9 Cr, 2.4 Ti, 1 Mo, the rest ---- Fe,

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ELECTRON BEAM PROCESSES



Figure 1. Scheme of ion-activated electron beam evaporation and condensation in vacuum: 1 — water-cooled crucibles with ingots being evaporated; 2 — ingot of base material; 3 — reaction gas supply; 4 — annular ionizer-electrode; 5 — electron beam for evaporation; 6 — vapor flows of materials being evaporated; 7 — substrate, on which acts bias potential; 8 — arc discharge; 9 — ingot with additive material

were used as the substrates. Roughness parameter of the specimens corresponded to $R_a = 1.8$ --2.2 µm. The specimens (from 30 to 60 pcs) were fixed in 1--2 rows in the immovable fitting-out above the evaporator, whereby the distance between the specimens and the pool surface in the crucible was 270 mm. Along the stationary located in the fitting-out specimens the temperature gradient within the range from 250 to 850 °C was created by the direct electron beam heating, due to which it became possible to produce for investigation in one experiment a set of specimens with the coatings, deposited at different values of the substrate temperature.

In the experiments at constant temperature of the substrate, in which conditions of deposition of the coatings on real blades were modeled, the specimens were fixed in the fitting-out, installed on the rotating horizontal shaft. Each specimen performed individual rotation in the vapor flow around its own axis.

Immediately before deposition of the coating in the vacuum chamber, surface of the heated specimens was cleaned by its bombardment by Ar^+ ions with the energy, approximately, 2--3 keV. Density of the ion flow, generated by the established inside the vacuum chamber at the distance 200 mm from the substrate gas-magnetron ion source, achieved, approximately, 1 mA/ cm², whereby rate of sputtering (etching) of surface of the specimens constituted 10 nm/ min.

Diameter of copper water-cooled crucibles for evaporation of the base material equaled 50 mm, and for evaporation of the additives ---- 25 mm. Titanium ingots of electron beam remelting of 49.5 and 24.5 mm diameter and small ingots of TiC or B_4C of 48 mm diameter, produced by the method of hot and cold pressing, were used as the material. Reaction of the TiN synthesis on the substrate was performed by simultaneous deposition of titanium and introduction into the working chamber of nitrogen at the pressure $2 \cdot 10^2$ Pa. At the initial instant of the titanium ingot evaporation the arc discharge was excited, and the binding layer of titanium of 2--4 µm thickness was deposited on the substrate. Then, without interrupting the titanium evaporation process, nitrogen was introduced into the working chamber and synthesis of the TiN coating of 10--20 µm thickness was performed at the condensation rate about 1 µm/min.

Investigation of the structure and determination of chemical composition of the coatings were performed under conditions of secondary and elastically repulsed electrons, using the CamScan 4D scanning electron microscope and energy-dispersion system of the INCA-200 Energy X-ray spectral microanalyser.

On this microscope microstructure of the coatings in cross section of the specimens was also analyzed.

Parameters of the crystalline lattice of the coatings were determined using the Dron-4 X-ray diffractometer in CuK_{α} radiation ($l_m = 0.154051$ nm).

Microhardness of the coatings in cross section and on the surface was measured using the microhardness gauge-attachment to the Polyvar Met microscope at the load 20 and 50 g on the basis of the results of 5--7 measurements with estimation of the mean-rootsquare error for the probability 0.76.

Adhesion tests were carried out by bending of the specimens with application of the three-point system of loading on the 12 mm basis. Length of the specimens equaled 22 mm, width ---- 6 mm, thickness ---- 2.4 mm. Rate of loading constituted 1 mm/ min. Minimal angle of bending of the specimen, at which first signs of failure and lamination of the coating appear, was used as criterion of the adhesion level. State of the coating after bending of the specimen was investigated on optical microscope at magnification power 400. Despite the fact that these tests do not give quantitative estimation of the level of adhesion of the coating to the substrate, the results obtained allowed comparing and qualitative estimating properties of different versions of the coatings.

Erosion resistance of the coatings was estimated in the abrasive wear tests in the flow of solid particles with application of the TsUK-3M installation according to the methodology presented in [12], by comparison of the wear (loss of the mass) of the coatings and the standard reference specimens.

Quartz sand, having size of the particles 200 μ m, was used as the abrasive material. Consumption of the abrasive was 125 g/min. The specimens were inspected and weighed after consumption of 2 kg of abrasive. To prevent influence of the quartz sand particle change in the process of the tests, the abrasive was not repeatedly used. Amount of the abrasive, consumed in one cycle of the tests, varied from 8 to 26 kg. The tests were carried out at the angles of the



Figure 2. Dependence of deposited titanium nitride microhardness upon temperature of stationary fixed substrate

abrasive flow attack of the surface of the specimens 30 and 90°, speed of the rotor rotation was 2350 rpm/min. Relative erosion resistance of the coatings was calculated according to the formula

$$\varepsilon = g_{\rm c} \rho_{\rm ref} / g_{\rm ref} \rho_{\rm c}$$

where g_c , g_{ref} are the average loss of the mass of the specimens with the coating and reference specimens; ρ_c , ρ_{ref} are the density of the coating and the reference material respectively.

In Figure 2 dependence of the deposited titanium nitride microhardness upon temperature of the stationary fixed substrate is presented. Monotonous growth of the TiN condensate microhardness from 19.0 to 23.6 GPa within the condensation temperature range from 200 to 500 °C was discovered. Further increase of the temperature did not cause microhardness increase of the condensed TiN layer.

In Figure 3 a typical microstructure of the erosion-resistant coating of 12 μ m thickness on the basis of titanium nitride, deposited on the rotating steel substrate at the temperature 460--490 °C and pressure in the chamber at the level 2 10^2 Pa, is presented. Identity of the coating microstructures on the basis of titanium nitride of 15--20 μ m thickness, deposited both on stationary and rotating substrates, should be noted.

Chemical composition of the TiN coatings (its stoichiometry) is determined by the amount of introduced into the vacuum chamber nitrogen. At optimal technological parameters of the process and tempera-



Figure 3. Microstructure of cross section of erosion-resistant titanium nitride base coating: 1 — substrate; 2 — binding titanium layer; 3 — TiN layer (×3000)



Figure 4. Fragmentation of erosion-resistant titanium nitride base coating after adhesion tests at bending angle of specimen 90°, and microstructure of cross section (×1200)

ture of the substrate within the range from 460 to 490 °C content of nitrogen in the TiN coating constitutes 20–23 wt.%, microhardness achieves 23.5 GPa, parameter of the deposited titanium nitride lattice was a = 0.4230 nm, which corresponds to that of the TiN coatings produced by traditional plasma-chemical methods [13, 14]. Investigation of the fine structure of the deposited coatings was not the goal of this investigation, but such characteristic peculiarity of the condensed titanium nitride microstructure as its nanocrystalline structure with size of the grains less than 30 nm should be noted.

Adhesion tests showed that the TiN coating, produced within the condensation temperature range 450--600 °C, did not laminate when the specimens were bent at 90°, only its cracking and fragmentation were detected (Figure 4). First cracks in the coating were visually registered at bending angles 7--12°.

Relative increase of the erosion resistance (Figure 5) of the specimens with the TiN coating, deposited on the rotating substrates, constitutes, in comparison with the base material, from 25 to 8 (at the test angles 30 and 90°, respectively).

Further improvement and development of the electron beam vapor-phase technology of deposition of the nitride-base coatings are possible by development of both simple (TiN, ZrN, CrN, etc.) and complex (TiAlN, TiCN, ZrCN) coatings, which may be graded, microlayer, or nanostructure ones. They are produced by change of the reactive gas pressure, allying of the



Figure 5. Change of mass of steel specimens without coating (reference ones) and with TiN coating of 15 μ m thickness in process of erosion resistance investigation at angle of attack 90°: 1 — steel; 2 — TiN; m_a — mass of abrasive; m_1 — loss of mass

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Figure 6. Microhardness of TiN (1) and TiN with addition of 3–5 wt.% Al_2O_3 (2) coatings depending upon substrate temperature

main evaporated ingot, application of the composite ingot, or introduction of the additive into the vapor flow by means of evaporation from the auxiliary crucible. For example, due to introduction of the aluminium oxide and aluminium (3--5 wt.%) into vapor flow in ion-activated electron beam evaporation of titanium microhardness of the deposited TiN-base coatings increased up to 26--28 GPa within the substrate temperature range 450--850 °C (Figure 6).

In Figure 7 structure of the microlayer TiN coating, containing interlayers of titanium, is presented. The coating is produced by periodic change of the amount of supplied nitrogen in the process of continuous evaporation of titanium. Erosion resistance of this coating increased 1.7 times despite reduction of its microhardness down to 20 GPa.

TiC-base coatings are the most distributed among the carbide ones. It is possible to produce by the method of direct electron beam evaporation of carbides hard TiC-base coatings at the substrate temperature above 700 °C [15]. At the condensation temperature 700 °C hardness of such coatings equals, approximately, 20 GPa and increases up to 24--25 GPa as temperature of the substrate increases up to 1300 °C. Below the condensation temperature 700 °C in the TiC coatings titanium and carbon in free state are also detected.

For better adhesion of the TiC-base coatings to the substrate primarily a binding layer of titanium of 3--4 μ m thickness was deposited on its surface. For this purpose from the second crucible of 25 mm diameter titanium was evaporated, the vapor flow being



Figure 7. Microstructure of cross section of microlayered erosionresistant titanium nitride base coating, containing interlayers enriched by titanium: 1 --- substrate; 2 --- binding layer; 3 --- titanium interlayer (×3000)



Figure 8. Microhardness of TiC coatings produced by direct (1) and ion-activated (2) electron beam evaporation

partially ionized by the arc discharge, and deposited on the substrate, which was under bias potential about 130 V. Then evaporation of titanium from the small crucible was stopped and practically without any interval evaporation of a small TiC ingot from the crucible of 50 mm diameter was performed.

Two versions of evaporation of a small TiC ingot from the crucible of 50 mm diameter were performed: direct and ion-activated evaporation with simultaneous ionization of the vapor flow by the arc discharge (arc current ---- 25 A, voltage ---- 45 V) and bias on the substrate (bias voltage was 130 V).

The rate of deposition was $1.2-1.5 \,\mu\text{m}/\text{min}$, thickness of the coating achieved $19-23 \,\mu\text{m}$. Chemical composition of the coatings corresponded to mean values of carbon and titanium concentrations in the ingot being evaporated; a small amount of oxygen was noted in the coating.

In Figure 8 dependences of microhardness of the coating surfaces, produced by direct and ion-activated electron beam evaporation of TiC, are compared. Ionization of vapor flow of non-stoichiometric composition and acceleration of positive ions activated process of TiC synthesis on the surface of condensation within the deposition temperature range 150--750 °C, whereby microhardness of the coatings, produced at the temperature 450--700 °C, remained constant and equaled (21 ± 1) GPa. In adhesion tests of steel and titanium (VT1 alloy) specimens with the TiC coating lamination of the coating was not registered up to the bending angles 90°.

In Figure 9 microstructure and distribution of the main chemical elements of one of the versions of the graded MgO-TiC damping coating of 26 μ m thickness, produced by evaporation of MgO and TiC ingots from separate crucibles at the substrate temperature 500 °C, are shown. Microhardness of the outer erosion-resistant layer is 18 GPa.

Into the list of the most promising materials, used as erosion-resistant coating, should be included boron carbide due to its high microhardness (more than 40 GPa) and low specific weight (2.52 g/cm^3) [16]. For deposition of boron carbide as the coating of different designations of up to 5 µm thickness ion-plasma technology, chemical deposition from the gas medium, arc and magnetron sputtering are used [17--20].



Figure 9. Distribution of chemical elements (a), and microstructure of cross section (b) of MgO-TiC graded coating (×2500)

The main disadvantages of the technologies of application of coatings from boron carbide remain to be low rates of deposition and occurrence of significant internal stresses by means of growth of the coating thickness. Plasma spraying ensures production of the coatings of more than 10 μ m thickness at increased rates, but internal porosity significantly limits their use as erosion- resistant ones [19]. Character of the mechanisms of the damage and failure of the coatings from boron carbide in the process of erosion tests proves the need to increase their fracture toughness [21].

Fine-crystalline two-phase metal, metal-ceramic and ceramic materials on the basis of hard and brittle phases A, containing 20--30 wt.% of the phase B with lower elastic and shear modules in comparison with the phase A, have characteristic maximums of mechanical strength, elasticity, and fracture toughness $K_{\rm Ic}$ properties [22].

In the PWI International Center for Electron Beam Technologies the technology for electron beam evaporation of the composite ingot from boron carbide and titanium pellet, located in upper part of the ingot, for deposition of multiphase coatings, in which mentioned mechanism for improvement of the mechanical properties is implemented, was developed.

In Figure 10, *a* microstructure of the multiphase graded boron carbide base coating of 15 μ m thickness, deposited on the specimens from stainless steel, is presented. Temperature of the specimens in the process of condensation did not exceed 500 °C, rate of

condensation was 1 μ m/min. Microhardness of the produced coating equaled 30 GPa. Investigations of the coating fine structure (Figure 10, *b*) allowed establishing that the boron carbide base layer represented, due to the peculiarity of the developed technology for evaporation and deposition, a multiphase nanostructure material with inclusions of the nanoparticles of about 10 nm size. It was established in adhesion tests of the boron carbide base coating of 10–15 μ m thickness that first cracks in the coating were visually detected at bending angles about 7°; lamination of the coating was not detected within the whole range of the bending angles up to 90°.

Erosion resistance of mentioned coatings of 10-15 μ m thickness on the specimens exceeded more than 20--50 times resistance of specimens without the coating at test angles of 90 and 30° respectively (Figure 11).

Presented primary results of the developments and investigations prove possibility of essential increase of the fields of application of the electron beam equipment and evaporation technology and are the basis for development of the graded erosion-resistant coatings (or erosion-resistant layers of the graded coatings) with high parameters of functional characteristics.

At present the PWI EBT IC proposes to the potential buyers its production capacities for electron beam deposition of hard erosion-resistant graded coatings and granting of licenses on application of the developed know-how in this field.



Figure 10. Microstructure of graded boron carbide base coating: - cross section (×3000); b ---- electron-microscopic image of fine structure (×2,000,000)

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Figure 11. Change of mass of steel specimens without coating (reference ones) and with graded $B_4C\text{-}base$ coating of 15 μm thickness in process of investigation of erosion resistance at angle of attack 90°: 1 ---- steel; 2 ---- B₄C

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MECHANICAL PROPERTIES AND STRUCTURE OF BRAZED JOINTS OF JS26VI CAST NICKEL ALLOY Part 2^{*}

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Interconnection of physical-chemical properties of the seam metal with microstructure of brazed joints of the JS26VI cast nickel alloy produced with application of complex brazing alloys, containing boron and silicon as depressants, is considered. Silicon was added in the form of a powder of the NS12 (Ni–12 wt.% Si) commercial brazing alloy. The best results were achieved in case of introduction into the complex brazing alloy of 20 wt.% NS12. In this case brazed joints were characterized by strength at the level 670–760 MPa, and their relative elongation achieved 13 % at room temperature.

Keywords: brazing in vacuum, JS26VI cast alloy, complex silicon, boron-containing brazing alloy, brazed joint, strength, elongation, structure, fracture

Cast heat-resistant alloys, used in industrial turbines and gas-turbine engines, are subjected to rather complex heat treatment for the purpose of increasing heterogeneity of the structure and strengthening and improvement of the metal plasticity. Three- or four-stage heat treatment is used for the blades both for homogenization of the cast structure and for improvement of the technological characteristics.

In case of repair brazing of the components from cast nickel heat-resistant alloys (HA) the same methods of treatment are used for renovation of the brazed joints (BJ), which include high-temperature annealing for equalization of chemical gradients, ensuring of full diffusion interaction between the solidified melt of the brazing alloy and walls of the item being brazed, and achievement of necessary mechanical characteristics of the seam metal and BJ in the process of operation [13]. These problems are in close interconnection with peculiarities of the BJ microstructure.

Metallography. Optimized conditions of brazing of JS26VI alloy (1220 °C, 15--20 min) ensured formation of braze seams of satisfactory quality. The alloy is characterized by high heat stability and, therefore, limited diffusion of the components. That's why all braze seams on the specimens had clearly outlined appearance with insignificantly developed zones of mutual diffusion (Figure 6). Majority of BJ, produced by the brazing alloy #1 + 60 % Rene-142 with additions of different fillers and the NS12 brazing alloy, had dense structure with minimal amount of defects. Width of the seams, produced by the contact brazing method, was 150--200 µm. At first stage the main task of the work consisted in selection of optimal composition of the complex brazing alloy, whereby both weight share 15-20 % of silicon-containing component of the NS12 brazing alloy (Figure 6, *a*, *c*) and ratio of components in the filler varied, using for this purpose either completely powder of the Rene-142 alloy or substituting part of it by the JS6U alloy powder (Figure 6, *g*). The optimal version of the complex brazing alloy was selected on the basis of the mechanical test results of the BJ specimens.

Results of the microstructure investigations allowed substantiating rational system of the brazing alloy. The main criterion for assessment of the seam quality was degree of its homogeneity in presence of minimal amount of course carbide phases and lowmelting eutectics. Selected versions of the brazing alloys #1 + Rene-142 and #1 + 20 % NS12 + Rene-142ensured high level of mechanical properties and respective structure of BJ.

Heat stability of the alloy at brazing temperature 1220 °C stipulates a limited interaction of the brazing alloy melt and the base being brazed. Only due to two-stage heat treatment a minimal redistribution of the alloying components, mainly tungsten, rhenium, vanadium and carbon, at the interphase boundary was detected (Figure 7, Table 3).

At second stage conditions of BJ heat treatment for their correspondence to the heat treatment conditions of the base alloy were corrected.

Leveling of the BJ microstructure is possible only at a rather high temperature of annealing (1160 °C), which «enters into contradiction» with structure of the seam metal and the fusion line, because during annealing coagulation of the strengthening γ' -phase particles (diameter of the particles achieves 2.0-

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Figure 6. Microstructure (×200) of braze seams formed on JS26VI alloy at 1220 °C for 15 min with application of different versions of composite brazing alloys: *a* — 25 % #1 + 15 % NS12 + 60 % Rene-142; *b*-*d* — 20 % #1 + 20 % NS12 + 60 % Rene-142; *e*, *f*, *h* — 40 % #1 + 60 % Rene-142; *g* — 40 % #1 + 30 % Rene-142 + 30 % JS6U (×270)

 $2.5~\mu m)$ occurs, and tensile strength of BJ significantly reduces at the test temperature 900--1000 °C.

That's why after homogenization, which allows causing cross diffusion of the components from the brazing alloy into the base and vise versa, it is necessary to introduce technological operation of ageing for precipitation of the secondary strengthening γ -phase. Two versions of ageing (1050 °C, 2 h and

900 °C, 3 h) for both brazing alloys ---- #1 + 60 %Rene-142 and #1 + 20 % NS12 + 60 % Rene-142 (Figure 6, *b*, *d*, *f*, *h*) were considered. Ageing temperature effected structure of the solidified seam metal and development of the BJ diffusion zone.

Full picture of structural changes, caused by stageby-stage heat treatment after brazing, is presented in Figures 8 and 9 and Table 4.

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Figure 7. Areas of X-ray spectral microanalysis of BJ formed at 1220 °C for 15 min and annealed at 1160 °C, 2 h + 900 °C, 3 h with application of two systems of brazing alloys (Table 3): a --- #1 + 20 % NS12 + 60 % Rene-142 (×180); b --- #1 + 60 % Rene-142 (×220)

In the post-brazing state at maximum possible rate of cooling 15--17 °C/ min the brazing alloy metal represents a multiphase system (see Figures 8 and 9). Depressants in the form of boron and silicon, which enter into composition of the brazing alloys, reduce melting point of the melt and temperature of full dissolution of the strengthening γ -phase, causing change of morphology of carbide reactions in the melt. At the temperature of full dissolution $T_{\rm f,d}$ of γ -phase redistribution of the alloying components between γ and γ -phases takes place in the brazing alloy metal, and, as a result, processes of carbide formation are intensified.

That's why during cooling with the furnace of brazed billets within temperature range 1220--1050 °C at a relatively low rate of solidification, decomposition of the primary striated carbides MeC with precipitation of the Me₆C carbides of eutectic type on the basis of tungsten, chromium, rhenium and molybdenum takes place. These carbides prevail in structure of the solidified seam metal after brazing as well as dispersed equiaxial carbides (Ta, Hf, Nb)C (Figure 9,

Table 3. Content of components in BJ JS26/#1 + 60 % Rene-142/JS26 without and with 20 % NS12 after annealing at 1160 °C, 2 h + 900 °C, 3 h (see Figure 7)

Investigated	No. of					V	Veight sh	are of con	nponents	, %					
zones of BJ	spectrum	Ni	Al	Co	Ti	Cr	W	Мо	Nb	С	v	Та	В	Re	Hf
			J	S26/#1	+ 20 %	NS12 -	+ 60 %	Rene-14	2/JS26	i					
JS26	Composition of alloy (regular)	Base	5.5 6.2	8.0 10.0	0.8 1.2	4.3 5.6	10.9 12.5	0.8 1.4	1.4 1.8	0.13 0.18	0.8 1.2				
Base metal	2	60.65	5.16	9.1	1.44	5.04	14.16	0.88	2.48		1.09				
Brazed joint metal	3	63.64	5.31	8.93	0.47	5.48	15.42				0.74				
#1 +	Composition of brazing	Base	3.2 3.8	8.8 9.5		13.0 14.0									
Rene-142 +	alloy (regular)		5.9 6.3	11.5 12.0	0.02	6.6 7.0	4.7 5.1	4.7 5.1	0.1	0.10 0.14		6.5	2.5	2.3 4.5	1.5
NS12													12Si		
Seam metal	4	68.24	4.31	8.82		6.74	3.91	0.88	2.93			2.9	2.78Si	0.82	0.56
				JS	26/#1	+ 60 %	Rene-14	12/JS26	6						
Base metal	1	63.54	5.56	9.05	0.91	5.06	13.47	0.59	0.90		0.93				
Brazed joint metal	2	62.92	5.44	9.81	0.41	6.47	12.09	1.03	0.62		0.43		-	0.79	
#1 +	Composition of brazing	Base	3.2 3.8	8.8 9.5		13.0 14.0							-		
Rene-142	alloy (regular)		5.9 6.3	11.5 12.0	0.02	6.6 7.0	4.7 5.1	4.7 5.1	0.10	0.10 0.14		6.5	2.5	2.3 4.5	1.5
Seam metal	3	62.63	5.43	10.1	0.34	9.25	5.83	1.0	0.68			3.5		1.23	





Figure 8. Pattern of structural changes in solidified metal of seams formed with application of brazing alloys #1 + 60 % Rene-142 with (*a*, *c*, *e*) and without (*b*, *d*, *f*) addition of silicon at different stages of heat treatment after brazing: *a*, *b* — after brazing at 1220 °C for 20 min; *c*, *d* — after annealing at 1160 °C, 2 h + 1050 °C, 2 h; *e*, *f* — after annealing at 1160 °C, 2 h + 900 °C, 3 h (×500)

Table 4). Redistribution of the alloying components and carbide-formation reactions proceed in HA mainly at temperature $T \ge 1245$ °C, but in case of performance of the brazing process at temperature 1220--1225 °C silicon and boron, used as the depressants, reduce characteristic temperatures of the brazing alloy melting and activate segregation processes in the brazing alloy melt at lower values of temperature.

Ageing completes redistribution of the components in the seam metal at the interphase boundary and enables fuller dissolution of complex-alloyed eutectics in it (Table 5). In dynamics structure of BJ is improved at a lower temperature of annealing and increased ageing time. The least developed diffusion zone was formed after ageing at 900 °C, 3 h in case of the silicon-containing brazing alloy use (Figure 10).

Low-temperature ageing at 900 °C causes additional precipitation of secondary γ -phase in the seam matrix, which stipulates satisfactory level of BJ strength of JS26VI alloy. In Figures 8 and 9 influence of silicon on the seam metal structure is registered. Its small amount (2 wt.%) introduced into the seam matrix (in the form of 20 wt.% of NS12 brazing alloy), enables refining of the grain and a fuller decomposition of the course carbide phases, precipitation of complex-alloyed eutectics in the inter-axial volumes of dendrites being excluded. Silicon disperses precipitated carbide eutectics, and homogenizing double-stage annealing enables further practically full dissolution of the carbide phase. That's why after low-temperature ageing the main part of the BJ metal constitutes a high-alloyed solution and a dispersed strengthening γ -phase.

While initially silicon is distributed mainly over grain boundaries (up to 5 %), after heat treatment its content equalizes and constitutes 2--3 % in the seam matrix. In case of presence in the brazing alloy of another depressant (boron), silicon is located mainly in central part of the seam, does not diffuse to the interphase boundary with the base, and does not form brittle silicides along the fusion line [14].





Figure 9. Areas of X-ray spectral microanalysis of separate phases constituting BJ seam metal #1 + 20 % NS12 + 60 % Rene-142 (*a*, *c*, *e*) and #1 + 60 % Rene-142 (*b*, *d*, *f*) (see Table 4) (×600): *a*, *b* --- after brazing; *c*, *d* --- after annealing at 1160 °C, 2 h + 1050 °C, 2 h; *e*, *f* --- after annealing at 1160 °C, 2 h + 900 °C, 3 h; *g* --- diffusion zone of BJ with 20 % NS12





Conditions of	No. of	Weight share of elements, %												
brazing and annealing	spectrum	С	Al	Si	Ti	Cr	Co	Ni	Nb	Мо	Hf	Та	w	Re
1220 °Ñ, 15 min	1	3.14	5.45	1.83		4.79	7.43	66.51		0.57		4.39	4.66	1.23
	2	3.20	3.59	5.59		6.07	9.54	72.00						
	3	2.54	2.98	5.17		4.81	8.59	70.64				3.71		1.56
	4	1.95	0.72	2.38	0.12	4.79	10.40	65.55	0.64		8.55	4.89		
	5	4.77				24.81	3.76	9.11		13.68		3.27	26.39	14.21
	6	4.68				28.84	3.41	7.18		13.52		3.20	23.58	15.59
	7	3.36	4.85	2.69		4.79	7.21	68.02				4.08	3.75	1.24
1220 °C,	1	2.16	2.64	3.12		7.78	11.17	69.09		0.46		1.43	1.13	1.04
15 min + 1160 °Ñ 2 h ⊥	2	1.41	4.42	3.27		5.40	7.99	68.15		0.81		3.53	2.69	2.34
1050 °Ñ, 2 h	3	3.44	0.64		0.26	3.10	10.13	61.50	0.83		12.53	6.82	0.74	
	4	11.89	0.23		0.52	1.02	0.57	4.84	3.15	1.43	40.75	33.51	0.36	1.71
	5	3.75	0.20	9.84	0.29	1.73	3.96	40.00	2.10	1.81	22.85	13.47		
	6	5.16	0.18			34.81	5.52	9.43		12.94		3.70	14.79	13.48
	7	4.35				29.05	3.83	6.69		14.09		3.47	22.08	16.43
	8	5.29				36.99	5.34	8.25		11.42		2.84	13.38	16.49
1220 °C, 15 min +	1	1.03	4.15	2.53		6.87	8.57	68.00		0.85		1.81	4.24	1.96
1160 °C, 2 h +	2	1.40	4.37	2.49		7.06	8.74	68.25		0.72		2.24	3.08	1.65
300 N, 3 II	3	2.27				34.43	4.88	12.40		13.06		2.23	15.63	15.10
	4	2.81				2.96	5.21	13.68		14.63		9.21	45.68	5.81
	5	8.39			0.47	0.71	0.71	6.07	1.76	0.71	24.81	55.51		1.06
	6	1.92	3.91	3.09		6.85	9.62	69.27		0.48		2.31	2.55	

Table 4. Content of elements in structure components of BJ seam JS26/20 % #1 + 20 % NS12 + 60 % Rene-142/JS26 (see Figure 9, *a*, *c*, *e*)

In presence of silicon in the matrix $(\gamma - \gamma')$ -eutectic almost completely dissolves, and primary striated carbide phase disappears. Isolated carbides precipitate in a more favorable equiaxial form.

Another pattern of structural transformations was detected in case of the base brazing alloy composition #1 + Rene-142 without NS12. This brazing mixture has an increased content of chromium, cobalt and boron (as a depressant). Presence of a big amount of alloying elements, including components brought by the Rene-142 filler, enable increase of the brazing alloy melting point, and in solidification ---- significant multiphase nature of the seam metal structure (Figure 9, *b*, Table 6).

Due to solidification after brazing the structure of the seam metal is formed, consisting of a high-alloy matrix solution over axes of dendrites, boundary developed carbide phases of eutectic type, and complex-

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Table 5. Content of elements in structure components of BJ fusion zone JS26/20 % NS12 + 20 % #1 + 60 % Rene-142/JS26 after annealing at 1160 °C, 2 h + 900 °C, 3 h (see Figure 9, g)

No. of						Weight s	hare of ele	ments, %					
spectrum	С	Al	Si	Ti	V	Cr	Со	Ni	Nb	Мо	Та	W	Re
1	1.60	3.80	1.57			6.95	9.39	67.21		0.52	2.21	5.08	1.66
2	0.74	4.11	1.46			6.89	9.88	67.47		0.57	1.74	5.44	1.70
3	2.28	4.07			0.70	6.51	10.32	65.79			2.52	7.91	
4		5.13		0.60	1.19	5.50	9.43	65.09	1.02	1.45		10.58	
5	2.38	0.52				3.73	5.96	19.14		9.60	4.61	48.33	5.73
6	2.85	2.90		0.36	0.67	4.43	6.21	44.49	0.77	3.08	1.74	31.06	1.44
7	15.58			11.16	1.65	0.70	0.72	5.70	42.59	3.29		16.93	1.66
8	1.16	4.73		0.94	0.98	2.71	6.69	52.37	1.39	2.20		26.83	

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Conditions of	No of					Wei	ght share o	of element	s, %				
annealing	spectrum	С	Al	Ti	Cr	Со	Ni	Nb	Мо	Hf	Та	W	Re
1220 °Ñ, 15 min	1	2.64	5.60		8.37	11.00	63.32		0.55		2.24	4.00	2.29
	2	2.63	6.51		6.45	9.59	65.08		0.73		4.53	3.10	1.39
	3	2.31	2.00		8.24	13.45	62.10			4.41	5.54	1.22	0.72
	4	4.90			28.01	3.81	6.98		9.38		5.38	29.79	11.75
	5	4.56	4.37		10.75	10.92	56.73		1.61		2.67	4.95	3.43
	6	9.72		1.56	1.16	0.90	3.80	2.58	0.96	13.06	63.85	2.40	
	7	8.36		1.60	1.16	1.23	4.13	2.81	1.03	14.85	62.04	1.03	1.77
1220 °C,	1		1.55		8.43	13.87	63.11	0.50		5.40	5.71	1.32	
15 min + 1160 °N, 2 h + 1050 °N. 2 h	2	1.23	4.56		8.28	12.12	67.58		0.38		3.26	1.82	0.79
	3	3.06	0.25		26.13	3.62	6.24		10.94	1.09	3.72	35.95	9.01
	4	2.82	0.12		41.21	6.19	9.16		8.53		2.22	20.40	9.35
	5	2.59			46.85	4.85	4.74		5.09	0.69	1.11	12.89	21.20
	6	9.12		1.30	0.58	1.16	3.93	2.34		22.72	58.85		
1220 °C,	1	1.73	5.01	0.38	8.34	12.05	65.63	0.42	0.29		2.13	2.92	1.07
15 min + 1160 °C, 2 h + 900 °Ñ. 3 h	2	3.26	4.64	0.36	8.71	12.40	65.76		0.37		1.72	2.78	
,	3	5.49	0.28		25.57	3.79	6.66		10.61	1.03	3.03	35.02	8.48
	4	12.01		3.40	0.66	1.34	4.66	10.18		18.26	49.48		
	5	4.01	0.14	0.36	24.03	3.84	8.56	0.65	12.61		1.23	37.57	7.00
	6	3.95	0.98	0.36	4.47	12.13	59.26	0.71		13.27	3.78		1.09
	7	11.09		2.98	0.77	1.31	4.66	13.91		21.64	42.65	0.98	
	8	5.09	0.23		24.44	4.37	9.76		10.77		3.32	34.56	7.45

 Table 6. Content of elements in structural components of BJ seam JS26/40 % #1 + 60 % Rene-142/JS26 (Figure 9, b, d, f)



Figure 10. Structural changes at seam--base boundary in BJ produced with application of composite brazing alloy #1 + 60 % Rene-142 with (*a*-*c*) and without (*d*-*f*) addition of silicon on JS26 alloy with different content of carbon (see Table 5) (×500), *a*, *d* --- with minimal amount of carbon, after annealing at 1160 °C, 2 h + 1050 °C, 2 h; *b*, *e* --- with maximal amount of carbon, after annealing at 1160 °C, 2 h + 900 °C, 3 h





alloyed eutectics in the interphase areas. After solidification the eutectics occupy big volumes in the seam metal. Their formation is connected with seasoning of the alloy at the temperature above $T_{\rm f.d}$ of γ -phase and cooling at a rate above 10 °C/min [15]. In case of presence in the brazing alloy 40 % #1 + 60 % Rene-142 of the depressant in the form of boron, temperature $T_{\rm f.d}$ of γ -phase reduces (from 1245 °C) down to the temperature, corresponding to the brazing temperature 1220 °C. In the process of solidification at a cooling rate 15–17 °C/min, complex eutectics form after brazing, provided hafnium and course carbide particles Me₆C and MeC are present in the melt.

Chemical composition of the formed eutectics is as follows, wt.%: (13.5-13.9)Co--(1.6-2.0)Al--(8.2-8.4)Cr--(4.4-5.4)Hf--(5.5-5.7)Ta--(1.2-1.3)W (Table 6, Figure 9). Composition of inter-dendrite eutectic phases in braze seams is typical for different base alloys (JS26, VJL12U, JS6U, etc.) [see 9, 10 in Part 1]. In the spectrum of chemical composition of the complex-alloyed eutectics the boron line is fixed. This shows that due to application of the brazing alloy #1 + 60 % Rene-42 boron, which acts as a depressant, does not form in small amounts up to ~1.0 % independent chromium or nickel borides, but enters into composition of the eutectics.

In the process of heat treatment (homogenization + ageing) composition of the seam metal equalizes, due to which complex-alloyed non-equilibrium eutectics resolve (Figure 8, Table 6).

In case of application of the composite brazing alloy #1 + 60 % Rene-142 (without silicon), fusion line of the base with the brazing alloy has a blurred unclear character, and the diffusion zone contains over the boundary in addition to the dispersed carbide phases equiaxial carbides of Me₆C type (Figure 10, *d*, *f*). A more blurred contour of the BJ boundary is explained by presence in the brazing alloy of boron with high diffusion factor, which due to its small atomic radius easily penetrates over grain boundaries into the thermostable zone.

Base metal. In the course of fulfillment of this work, connected with formation of BJ, heat treatment and investigation of mechanical properties, divergences occurred both in structure of BJ and values of their short-term strength and plasticity. The main regularity of the results obtained preserved and consisted in stable increase of short-term strength and plasticity of BJ in case of introduction into the base brazing alloy (#1 + 60 % Rene-142) of 20 % NS12 powder (Ni--12 % Si). Silicon exerted favorable action on structure of the braze seam being formed: refined the grain and accelerated progress of carbide transformations in the seam system, preventing formation of complex-alloyed boundary eutectics. Cast billets of the JS26VI alloy, supplied from different plants, were used in the work.

Under the same conditions of BJ production (brazing at 1220 °C, 20 min) and standard heat treatment (annealing at 1160 °C, 2 h + 1050 °C, 2 h) at the second stage of investigations reduction of the BJ strength and plasticity was registered. Investigation of the microstructure showed differences in width and development of the fusion zone (diffusion zone) in the base metal (Figure 11).

In the other case precipitation of high amount of acicular carbide phase occurred at the seam--base boundary in the process of subsequent heat treatment, whereby width of the diffusion zone increased up to 30-- $50 \mu m$ and more.

Metallographic analysis of structure of the BJ seams explains this phenomenon in the following way. Mechanical properties of the joints directly depended upon properties of the base metal. Structure of the base metal differed, which determined difference in mechanical properties of BJ. At high content of carbon (0.15-0.16%) in the base a developed diffusion zone, containing high amount of carbide phases, is formed in the process of the BJ formation at the seam boundary (Figure 6, *b*, *d*, *f*, *h*).

Prior to brazing the billets were annealed in vacuum at 1210--1220 °C within 1 h. Such heat treatment minimally affected initial structure of the base. At the same time detailed analysis of JS26VI alloy showed presence of a different structure of the metal in two cases. At the first stage of the investigations structure of the alloy had a clearly pronounced dendrite character with inhomogeneity of γ -phase over axes and inter-dendrite areas (where boundary (γ - γ)eutectics occurred) and presence of primary skeletonlike (Ti, Nb)C and isolated equiaxial substitution (Ti, Nb, W)C carbides. Such state corresponded to JS26VI alloy after solidification.

In the second case a changed structure with a higher volume share of carbide phases of different type was detected in the alloy. Here boundary $(\gamma - \gamma)$ -eutectics were almost completely absent, which is characteristic of the JS26 alloy with weight share of carbon at maximally possible level (C = 0.16 %). Carbide phases are represented by three modifications: equiaxial (Ti, Nb, W)C one, acicular tungstenbase Me₆C carbides, and Me₂₃C₆ carbides. The latter are formed inside MeC carbides and are the result of the progress of low-temperature transformation at long-term ageing.

So, comparative phase analysis of the JS26VI alloy structure of different supplies confirmed that in first case the alloy had a reduced amount of carbon (0.10– 0.14 %) and was in the post-hardening state, and in second case it had a maximally possible content of carbon and was subjected to full cycle of heat treatment (homogenization and stepwise ageing). Values of the BJ properties in the alloy with a minimal content of carbon were higher than in the alloy with a maximally possible content of this element. Formation (in the second case) of acicular Me₆C carbides at the base--seam boundary exerts effect of incision in the tests and causes reduction of the BJ plasticity.

Homogenization annealing of BJ at 1160 °C with subsequent ageing at 1050 and 900 °C allowed getting





Figure 11. Differences in $(\gamma - \gamma)$ -microstructure of seam metal (a, b), diffusion zone (c, d) and base alloy (e, f) in BJ 20 % #1 + 20 % NS12 + 60 % Rene-142/JS26 after different kinds of heat treatment (×5000); *a, c, e* — after annealing at 1160 °C, 2 h + 1050 °C, 2 h; *b, d, f* — after annealing at 1160 °C, 2 h + 900 °C, 3 h

uniformly distributed in the seam matrix strengthening γ -phase of 0.3--0.5 μ m size, which is an efficient structural factor that increases creep resistance.

Ageing at the temperature 900 °C enables fuller precipitation from the solution of the strengthening γ -phase [16], i.e. increase of its volume share (Figure 11). At low-temperature ageing width of the BJ diffusion zone preserves, whereby carbide phases have a smaller size and are partially globulized.

In the post-brazing state (quasi-hardening) the base metal has a structure with a higher volume share (up to 70 %) of the cubic γ -phase with average size 0.5 µm. However, conditions of the BJ transformation into the equilibrium state require for performance of heat treatment, which causes partial dissolution and significant growth of γ -phase (Figure 12, *c*, *e*). Subsequent stage of ageing ensures precipitation of the secondary disperse phase from the oversaturated solid solution, determining combination of high values of

strength and plastic characteristics, which was confirmed by the mechanical test results.

Ageing at 900 °C for 3--4 h enables formation of a structure with distribution of the parts of different dispersity (course of 0.9--1.5 μ m, and fine of 0.12--0.18 μ m). After such ageing the alloy has maximum value of σ_t and sensitivity to the notch weaker, because the morphology of both carbide phases and γ -phase changes and, therefore, residual plasticity of BJ increases.

In Table 7 results of measurement of microhardness, which characterize structural changes occurring in two selected systems of brazing alloys at different stages of the BJ heat treatment, are generalized. Microhardness increased at low-temperature ageing because of precipitation of the disperse secondary γ phase both in the diffusion zone of the seam metal matrix and in the base. As ageing temperature reduced and duration of seasoning at 900 °C increased, partial





Figure 12. Change of strengthening γ -phase morphology in JS26 alloy of equiaxial crystallization (*a*, *c*, *e*) and at boundary with braze seam (*b*, *d*, *f*) after different stages of heat treatment (×500): *a*, *b* — hardening (brazing); *c*, *d* — homogenization + ageing; *e*, *f* — ageing

Table 7. Microhardness of separate structural components of BJ produced with application of two main systems of brazing alloys in post-brazing state and after different stages of heat treatment

	20 % #1 +	20 % NS12 + 60 %	6 Rene-142	40 % #1 + 60 % Rene-142				
BJ structural components	Initial	1160 °C, 2 h + 1050 °Ñ, 2 h	1050 °Ñ, 2 h + 900 °Ñ, 2 h	Initial	1160 °Ñ, 2 h + 1050 °Ñ, 2 h	1160 °Ñ, 2 h + 900 °Ñ, 3 h		
Base alloy	3660	4200	4240	3950	4260	4240		
Diffusion zone	4130	4885	4655	4155	5190	4980		
Seam matrix	4540	4320	4265	3800	4440	4255		
High-temperature eutectics	7380		Placer	8250	8230	8730		
Intermetallics on basis of Ni_xHf_y	-	6585	6250		-			
Double Me ₆ C carbides	Eutectic of 21730 type	More dispersed 21885	21265	Eutectic of 20645 type	Course 20985	21925		
Chrome carbides of Cr ₂₃ C ₆ type		13180	Very fine		12985	13150		







Figure 13. Microstructure of braze seams in longitudinal (a) and cross (b) sections on blades from JS26NK alloy produced with application of composite brazing alloy #1 + 20 % NS12 + 60 % Rene-142 at 1220 °C for 20 min in model gaps of 350 (a) and 525 (b) μ m width

dissolution and coagulation of the acicular phases in the diffusion zone occurred, which caused reduction of microhardness at the interphase boundary.

Increase of the ageing time up to 16 h at the reduced temperature (870 °C) creates possibility for a fuller precipitation from the solution of secondary γ -phase, decomposition and spheroidization of the carbide phases, and, as a result, enhancement of the BJ long-term strength at high temperature.

Developed technology of brazing of plates was used for repair of model blades of the aviation gasturbine engine, fabricated from the JS26 alloy. Absence of cross boundaries of the crystallites and availability of the pronounced direction of the dendrite growth in the metal structure along the blade axis allowed assuming that they were manufactured by directional solidification (DS). Gaps of 350, 530 and 950 μ m width and up to 10 mm length were made in the components by the electric erosion cutting method. After respective preparation of the blades for brazing, including blowing of the channel surface by SiC powder and vacuum annealing, the base brazing mixture (40 % #1 + 60 % Rene-142) and the mixture containing additionally 15--20 wt.% NS12, were applied on the gaps from the side of the pan. Brazing was performed at 1220 °C within 20 min.

Brazing alloy, applied from the side of the pan, melted, flowed into the gap, penetrated on the rear surface (back), and wetted it. Beads of brazing alloys above the gap, as well as the braze seams themselves, had a dense homogeneous microstructure. Complete filling of the channels at the depth up to 4 mm was achieved. In case of use of the silicon-containing brazing alloys a noticeable spreading of the brazing alloy melt over back of the blades was registered.

For leveling metal structure of the seams the blades were annealed after brazing at 1160 °C, 2 h with subsequent ageing at 1050 °C, 2 h. Heat treatment ensured full dissolution of boundary carbides of the type of course striated phases and complex-alloyed eutectics. At volume content 32–28 % size of the γ -phase particles in the seam metal constituted 0.2--0.6 µm, and in the base metal of the blade ---- 0.5--1.5 µm.

Redistribution of the alloying elements in the seam matrix and dispersion (full dissolution of course carbide particles in the matrix volume) positively affect both strength and plastic properties of the BJ metal.



Conditions of brazing and rational compositions of the brazing alloys allowed forming in the gaps defect-free braze seams, metal microstructure of which in longitudinal and cross directions is presented in Figure 13.

We have detected insignificant re-crystallization of the near-surface layer metal of the blade fur-tree root from JS26NK alloy, which could be connected with conditions of the blade root threading [17], and insignificant changes in the surface chemistry, more pronounced in the proximity of the tooth apex, but not in the recess. General thickness of the layer with changed chemistry constituted 5--6 μ m.

Surface layer of the fir-tree root metal is impoverished in regard to aluminium and chromium within 1.0--1.2 and up to 2 wt.%, respectively. In the near-surface aluminium-impoverished zone discrete particles of secondary strengthening γ -phase (Ni₃Al), precipitated during ageing and in the process of cooling, were detected. According to the layer-by-layer microhardness measurement data, loss of strength of the blade root metal from JS26NK alloy after isothermal brazing at 1220--1230 °C within 20 min did not occur.

CONCLUSIONS

1. Experimental-commercial technology for repair brazing of components, made from cast JS26VI nickel alloy of equiaxial crystallization, is developed. At 20 °C stable level of strength (σ_t = 700--730 MPa) and plasticity (ϵ = 6.5--13 %) of this alloy BJ, produced by the method of isothermal brazing at 1220 °C within 15--20 min in vacuum (5--8) $\cdot 10^{-3}$ Pa after two-stage heat treatment, are achieved.

2. Statistically processed tensile test results of BJ, produced by different brazing alloys according to two conditions of brazing at 1220 and 1230 °C, are presented.

3. It is established that upper threshold temperature of the JS26VI alloy brazing by Ni--Co--Cr--Al--2.5 % B (#1) base brazing alloy and complex boronand silicon-containing brazing alloy with addition of 20 % NS12 is 1230 °C, at which brittle failure of BJ after achievement of yield strength occurs.

4. It is shown that real strength of the BJ seam metal of JS26VI alloy exceeds that of the base metal by 150--180 MPa. This is confirmed by test results of

BJ of VJL12U and JS6U nickel alloys, produced by the same brazing alloys and subjected to a similar heat treatment. Scatter of mechanical property values of BJ of JS26 alloy is connected with structural state of the cast metal, content of carbon in it, and different crystallographic orientation of the growth dendrites in the joint plates being connected.

5. It is determined that silicon, introduced into boron-containing brazing alloy (#1) in the form of powder of Ni-2 % Si brazing alloy, increases technological plasticity of BJ of JS26 alloy at 20 °C due to refining of the precipitated carbide phases, but reduces short- and long-term strength at the temperature values 900-1000 °C. BJ, produced with application of the complex brazing alloy with silicon, are less durable at high temperature in comparison with BJ, formed just by Ni--Co--Cr--Al--2.5 % B (#1) brazing alloy.

6. It is established that satisfactory level of hightemperature strength of BJ of JS26VI alloy is achieved due to application of optimal conditions of heat treatment. Ageing at 900 °C, 3--4 h after homogenization annealing at 1160 °C ensures two-fold increase of short- and long-term strength of BJ at 900 °C.

7. Experimental-commercial technology for repair brazing of blades from JS26VI alloy with application of silicon- and boron-containing brazing alloy with filler from Rene-42 alloy is developed. Model gaps of 350, 525 and 950 μ m width were used in the blades. The gap width did not exert noticeable influence on quality of the braze seam, formed at 1220 °C within 20 min. Significant impoverishment of surface of the fur-tree root metal during vacuum brazing and final heat treatment of renovated in vacuum blades does not occur.

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GENERAL PROBLEMS OF METALLURGY



PHYSICAL-CHEMICAL PROCESSES IN COMPOSITE POWDER PARTICLES UNDER CONDITIONS OF PLASMA SPRAYING

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Results of investigation of composite powders produced by the SHS method and containing in its composition a metal component (Ni, NiAl, NiCrBSi) and inclusions of refractory compounds (titanium and chromium carbides, titanium silicide and aluminium oxide) are presented. Physical-chemical processes, occurring in particles of these composite powders in the course of their heating and movement in high-temperature gas flow, are investigated.

Keywords: composite powders, self-propagating high-temperature synthesis, plasma spraying, morphology, structure, phase composition

Protection of structural materials against wear at high temperatures is one of important problems of the stateof-the-art materials science. Most fully complex of requirements to the protection coatings can be met by development of the composite coatings, consisting of several components. The method of self-propagating high-temperature synthesis (SHS) ---- the process of the chemical reaction wave movement over the mixture of reagents with formation of solid final products ---- is characterized by the possibilities of producing composite powders (CP) with wide range of compositions. The SHS process represents conditions of passing of the exothermal reaction (reaction of burning), in which heat output is localized in the layer and is transferred from a layer to a layer as a result of heat transfer [1-5]. Chemical transformation in the SHS process is concentrated in the high-temperature zone, originated as a result of self-ignition due to high exothermicity of the synthesis process, and traveling within the volume at a significant linear speed due to heat transfer from the hot layers to the cold ones. The range of physical parameters, characterizing the process, is extremely wide (temperature ---- 300--4000 °C, linear speed ---0.1-30 cm/s, rate of the substance heating -1.10^2 - $1.10^{6} \text{ deg}/\text{ s}$ [4].

The SHS process, characterized by insignificant energy consumption, simplicity of the equipment, high speed and purity of the product attracted attention of the specialists. First of all it is connected with possibility of producing high-quality powders of refractory compounds (borides, carbides, nitrides, silicides, etc.) used as the basis for production of heat-resistant, superhard, superconducting, semiconducting, insulation, and other kinds of state-of-the-art materials. As theoretical and experimental investigation of SHS progressed, new and new possibilities of this process were discovered, and by now it acquired features of the independent scientific-technical field. To the fields of technology, in which the SHS process found its practical use, relates technology of the coating application by the spraying and surfacing methods in the following cases [6–10]:

• application of different initial powders, including the composite ones, produced by means of SHS and having certain advantages in comparison with the powders manufactured by other methods (composition, structure, cost, etc.);

• combination of the processes of coating application and synthesis of the coating materials by the SHS method, and use of the reaction heat for improving quality of the layer being formed.

For the thermal spraying (TS) many powders, produced by the SHS method, are of interest: titanium, tantalum, zirconium, hafnium nitrides, etc.; titanium, tungsten, titanium--chromium carbides, etc.; titanium and titanium--chromium borides; molybdenum silicides and sulfides, and different oxides. In addition, specificity of SHS ensures possibility of producing a number of CP, among which should be noted such compositions as chromium carbide--aluminium oxide, titanium carbide--ferrochromium, and chromium carbide--nichrome. Coatings from these powders are characterized by high wear-, heat- and thermal resistnce [7, 11, 12].

Fine mixing of the components in initial particles ensures homogeneity of their distribution within the coating volume. Examples of such compositions are Ti_5Si_3 --TiN, Cr_5Si_3 -- Cr_3C_2 , Al--TiC, TiC--TiSi_2, and $MoSi_2$ --SiC [7, 10--13]. High possibilities of the matrix composition and the inclusions control in SHS allow developing new materials for TS with a preset complex of properties.

Another aspect of the SHS method use in the TS technology consists in unification of the spraying processes and synthesis of the material being sprayed. For the first time this idea was implemented by the «Metco» company in application of the nickel--aluminium composition in the form of aluminium particles with a nickel coating [7]. Origination as a result of the exothermal reaction of the internal heat source

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allowed increasing thermal activity of the particles being sprayed and strength of the coating bond with the base.

To the compositions, in which interaction of the components proceeds with exothermal effect and can proceed in the SHS mode, relates combination of metals (*d*-elements of the IV, V and VI groups) with non-metal refractory compounds (SiC, Si₃N₄, B₄C, BN). Products of the interaction (mixtures of carbides and borides, carbides and silicides, silicides and nitrides, and borides and nitrides) are characterized by the complex of useful service properties, first of all, resistance against wear and corrosion.

Data on application of the SHS powders in the TS technology of coatings and production of coatings by unification of the spraying processes with synthesis of the spraying materials are presented in [7, 14--22]. Application of the SHS powders for TS creates possibility for forming coatings from the refractory compound--metal bond composition without application of mechanical mixtures and clad powders, whereby uniform distribution of components within the coating volume in comparison with the mechanical mixture is ensured because of presence of both components within the volume of each particle.

Physical-chemical processes, proceeding within volume of the CP particles during the period of its heating and movement in the high-temperature gas flow, are of great significance for formation of the coating. Under conditions of spraying of homogeneous materials surface of a particle serves as the interphase boundary. Gas environment and physical-chemical processes are mainly limited by interaction of the particle material with the environment (oxidation, nitration, reduction, etc.). Heterogeneity of the CP particle material composition stipulates origination of new interphase interfaces within the particle volume, due to which possibility of the progress of additional physical-chemical processes occurs.

At the same time, it is possible to assume that the following processes take place [23]:

• in case of occurrence of the melt of one of the components ---- wetting of surfaces of other components by it and, depending upon angle of wetting, spreading or coagulation into drops of the formed melt mass;

 interaction of the particle component melts with formation of new phases; • reaction interaction of the components with synthesis of new compounds, accompanied by a respective heat effect of the reaction process. In addition, the possibility is preserved of interaction with the environment ---- components of the gas flow and ambient atmosphere.

In this work structure, phase composition, and technological properties of CP, produced by the SHS method and containing in their composition metal matrix (Ni, NiAl, NiCrBSi), and inclusions of the refractory compounds: (titanium, chromium) carbides, titanium silicide, aluminium oxide, are investigated^{*}. In production of powders by the SHS method burning of the mixture charge of initial components proceeded in 30-litre reactors in inert gas. The SHS product represented a sintered shapeless mass, subjected to mechanical refining and classification on vibration screens for separation of particles of the necessary size (40--100 µm). In addition, physicalchemical processes, proceeding in the CP particles within the period of their heating and movements in high-temperature gas flow, were studied.

TiSi₂--Al₂O₃-base compositions were selected as objects of the investigation. In the titanium disilicide-aluminium oxide composition hardness ratio of the components constitutes 0.35--0.36, which corresponds to the requirements, established for erosion-resistant materials [24]. High resistance of the components against oxidation at 800--1000 °C allows using this material as a refractory one.

For producing coatings with designed properties introduction into the $TiSi_2$ -- Al_2O_3 -base CP composition of the following additional components was used: nickel for improvement of the coating plasticity; NiAl for increasing high-temperature strength; NiCrBSi and TiC for increasing wear resistance at temperature 600 °C; Cr_3C_2 for increasing wear resistance at the temperature values up to 800--900 °C.

In Table 1 composition of the CP, and in Table 2 main physical characteristics of CP and their components, are presented. Content of the metal bond in majority of cases constitutes 10–11, and in the composition with NiCrBSi ---- 28 vol.%.

The CP components significantly differ by level of the melting point (600--1000 °C) and density. Averaged by the mass value of density of the investigated powders constitutes 4.3--4.9 g/cm³, which exceeds 1.22--1.42 times density of the main refractory com-

Table 1. CP	composition
-------------	-------------

	Weight / volume share of components %										
Composition type	TiSi ₂	Al ₂ O ₃	NiCrBSi	$\tilde{N}r_3\tilde{N}_2$	Ni	TiÑ	NiAl				
TiSi2Al2O3NiCrBSi	35/43	20/28.2	45/28.8								
$TiSi_2$ Al_2O_3 $\tilde{N}r_3\tilde{N}_2$ Ni	37/39.4	40⁄49		3/1.9	20/9.7						
TiSi ₂ Al ₂ O ₃ TiCÑr ₃ Ñ ₂ NiAl	25/26.9	34/42		10/6.5		15/13.1	16/11.5				

^{*}The powders are produced in state scientific-technical center «Delta», Tbilisi, Georgia, under guidance of Prof. O. Okrostsvaridze.



ponent Al_2O_3 . In this connection one may assume that melting of Al_2O_3 in composition of the CP particles in plasma spraying can occur at lower energy values of the high-temperature gas flow than usually because of reduction of the movement speed of the particles in the jet and duration of the heating process [25].

In the course of investigations complex methodology was used, including metallography (optical microscope «Neophot-32» with an attachment for digital photographing); microdurometry (the LECO microhardness meter M-400 with automatic system of loading); scanning electron microscopy (JSM-840 scanning electron microscope). X-ray diffraction phase analysis (XDPA) was performed on the Dron-UM1 diffractometer in monochromatic CuK_{α} and CoK_{α} radiation.

Technological properties of powders (fluidity and bulk density) were determined according to GOST 20899--75 and GOST 19440--74.

Investigation of the morphology and structure of particles of the initial powders, produced by the SHS method, showed (Figure 1) that they had irregular shape and heterogeneous internal structure, representing a conglomerate of separate phase components. Phase composition of the SHS powders insignificantly differs from the calculated one; traces of oxides and/or unreacted residues of the initial charge were detected in them (Table 3). All powders had fluidity at the level of 25-45 s/ 50 g.

For analysis of the character of physical-chemical processes, proceeding under CP TS conditions, samples of the powders, which passed the heating and acceleration zone, were placed into water pool. Ware surface was located at the distance 200 mm from the nozzle edge. Spraying was performed on the UPU-8M plasma installation. Technological conditions of spraying corresponded to those of the plasma coating formation: I = 450-500 A; U = 40-55 V; flow rate of the plasma gas: argon ---- 27, nitrogen ---- 3.5 l/ min; spraying distance L = 120-200 mm; size of the particles was 80--40 µm.

Table	3.	CP	characteristics

 Table 2. Main physical characteristics of CP and their components

Material	Ò _m , ⁰Ñ	Density, g/cm^3
TiSi ₂ Al ₂ O ₃ **NiCrBSi*	10002050	~4.935***
$TiSi_2-Al_2O_3^{**}-Nr_3N_2-Ni^*$	14552050	4.286***
$TiSi_2$ Al_2O_3 TiC^{**} $\tilde{N}r_3\tilde{N}_2$ $NiAl^*$	16382780	4.32***
Al_2O_3	2050	3.50
TiSi ₂	1500	4.02
$\mathbf{\tilde{N}r_3\tilde{N}_2}$	1830	6.68
TiC	2780	4.92
Ni	1455	8.90
NiAl	1638	6.02
NiCrBSi	~1000	~7.70
*The most low-melting component	** The most ref	fractory compo-

nent. "Mean density of CP particle.

The TiSi₂--Al₂O₃--NiCrBSi powder (Figure 2, *a*-c), caught after plasma spraying, consists of the particles of two types ---- of irregular 1 and spherical 2, 3 shape. Surface of the particles of irregular shape is covered by small drops of the melt, formed as a result of the partial coagulation of the melt film on surface of the particle (Figure 2, *b*), and smooth areas. This is consequence of the interphase interaction within the particle volume of such components as TiSi₂ and Al₂O₃ at the stage of the TiSi₂ melting, Al₂O₃ being preserved in solid state.

Spherical 2 and orbicular particles in the form of two soldered hemispheres 3 represent, according to data of the analysis, the most low-melting (~1000 °C) component ---- NiCrBSi alloy, precipitating from the composite particle volume. In case of its partial alloying in interaction with TiSi₂ transformation of the sphere into the shape occurs, consisting of two spherical interconnected formations (a «nested doll») with separation of the volumes, containing different amount of titanium.

	Calculated phase	Bulk density	XDPA results [*]			
Powder composition	composition, wt.%	g/cm ³	Initial powders	Powders caught after plasma spraying		
TiSi ₂ Al ₂ O ₃ NiCrÂSi	~35 TiSi ₂ ~20 Al ₂ O ₃ ~45 NiCrBSi	2.61	$Ni_{16}Cr_6Si$, Ni, TiSi ₂ , α -Al ₂ O ₃ , B ₂ O ₃ traces	Ni, Ni ₁₆ Ti ₆ Si ₇ , γ -Al ₂ O ₃ , α -Al ₂ O ₃ , Ti ₅ Si ₃ , TiSi ₂ , NiCr ₂ O ₄		
$TiSi_2-Al_2O_3-Cr_3C_2-Ni$	~43 TiSi ₂ ~36 Al ₂ O ₃ ~5 Cr ₃ C ₂ ~16 Ni	3.09	TiSi ₂ , α-Al ₂ O ₃ , Ni, Cr ₃ C ₂ , Si traces	$\begin{array}{l} \gamma\text{-}Al_2O_3, \ \alpha\text{-}Al_2O_3, \ TiSi, \ Ni, \\ Cr_3\tilde{N}_2, \ Cr_7\tilde{N}_3, \ Ni_3Ti_2Si, \\ Ni \hat{I} traces \end{array}$		
TiSi ₂ Al ₂ O ₃ TiÑCr ₃ C ₂ NiAl	~17 TiSi ₂ ~25 Al ₂ O ₃ ~28 Cr ₃ C ₂ ~10 TiC ~20 NiAl	2.42	$\begin{array}{l} \alpha \text{-Al}_2 O_3, \ Cr_3 C_2, \ TiSi_2, \\ \text{TiC, NiAl, Ni}_3 Al, \\ Cr_7 C_3 \ traces \end{array}$	$\begin{array}{l} \gamma\text{-}Al_2O_3, \ \alpha\text{-}Al_2O_3, \ NiAl, \\ Cr_7C_3, \ TiSi, \ traces \ TiC, \\ Ni_2Al_{18}O_{29}, \ Ti_2Cr_2O_7 \end{array}$		

Phases are arranged according to reduction of intensity of the X-ray peaks.





Figure 1. Appearance (a, c, e) and microstructure (b, d, f) of initial CP particles: a, c ---- TiSi₂-Al₂O₃--NiCrBSi; c, d ---- TiSi₂-Al₂O₃--NiCrBSi; c, d ---- TiSi₂-Al₂O₃------ Cr₃C₂-NiAl; a, c, e ----- ×400; b, d ---- ×1000; f ---- ×800

So, in the process of heating of the particles in the plasma flow mainly division of the components of initial composition into two groups occurs: $TiSi_2$ --Al₂O₃ with local melting of $TiSi_2$, and NiCrBSi, which achieved full melting and spheroidization.

It was discovered in investigation of the powder $TiSi_2$ -- Al_2O_3 -- Cr_3C_2 --Ni particles (Figure 2, *d*-*f*) that in addition to the particles, which have signs of local melting of the surface 1, in the caught powder a significant number of spheroidal particles were present, which passed the phase of full melting. They have the shape of both correct spheres and «nested dolls» 3. In this case interphase interaction in two pairs of components $TiSi_2$ -- Al_2O_3 and Cr_3C_2 --Ni occurs. Presence of titanium enables improvement of the Al_2O_3 surface wettability, and spreading of nickel over Cr_3C_2 surface proceeds even more intensively. Contact angle of Cr_3C_2 wetting by nickel melt equals zero,

and interaction of nickel and Cr_3C_2 proceeds very quickly with formation of eutectic. The «nested dolls» contain product of both types of interaction. According to the analysis data, the «nested doll» consists of TiSi₂--Al₂O₃ (4) and Ni--Cr₃C₂ (5).

Spherical foundered particles represent the result of interaction of nickel and Cr_3C_2 ; particles with signs of local melting mainly consist of TiSi₂ and Al₂O₃.

Investigated TiSi₂--Al₂O₃--TiC--Cr₃C₂--NiAl powder contains the biggest number of refractory components (42 vol.% Al₂O₃, 13 vol.% TiC) in combination with low density ---- 4.32 g/cm³ (Tables 1 and 2). This hinders ensuring of sufficient preheating of the particles under conditions of plasma spraying.

Majority of caught from the plasma flow particles (Figure 2, g, h) have irregular shape with different degrees of local melting of the surface 1. Spherical particles 2 are, mainly, according to results of the

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Figure 2. Appearance of particles of composite powders caught from plasma flow: a-c --- TiSi₂-Al₂O₃--NiCrBSi; d-f --- TiSi₂-Al₂O₃--Cr₃C₂-Ni; g-i --- TiSi₂-Al₂O₃--TiC--Cr₃C₂-NiAl; a --- ×200; b --- ×800; c, f --- ×2000; d, g --- ×400; e, h --- ×950; i --- ×1800

analysis, the product of interaction of NiAl and Cr_3C_2 . The «nested dolls» combine within their volume products of interaction of TiSi₂--Al₂O₃ (3) and NiAl-- Cr_3C_2 --TiC (4).

Analysis of change of the shape and composition of the CP particles, caused by their passage through the high-temperature plasma flow, allows drawing conclusion that the main kind of the proceeding in this case physical-chemical processes is interphase interaction of the melts of their components with a solid surface (or melt) of another component. The reason of these changes consists in proceeding of the interphase interaction processes between components of the particles, melting point and density of which significantly differ (600--1000 °C).

Processes, which cause change of phase composition of the powders due to their heating in the plasma jet flow and hardening in the water pool, come to the following:

• in all investigated powders allotropic transformation α -Al₂O₃ $\rightarrow \gamma$ -Al₂O₃ occurs, which causes appearance of a new γ -phase;

• in the powders, containing $TiSi_2-Al_2O_3$ and more low-melting components (Ni, NiAl or NiCrBSi), separation of the initial composition components into two groups takes place: $TiSi_2-Al_2O_3$ with local melting

of ${\rm TiSi}_2,$ and nickel alloy, which achieved full melting and spheroidization;

• titanium disilicide transforms, as a result of partial loss of silicon, into lower silicides TiSi and Ti₅Si₃;

• in the powders, containing NiAl or NiCrBSi alloy, complex oxides $NiCr_2O_4$ and $Ni_2Al_{18}O_{29}$ form due to oxidation;

• in case of Cr_3C_2 chromium carbide oxidation, amount of the lowest Cr_7C_3 chromium carbide increases in the coatings because of burn-out of carbon;

• interphase interaction processes proceed between coating components with formation of new phases and their subsequent oxidation with formation of $Ti_2Cr_2O_7$, Ni_3Ti_2Si , $Ni_{16}Ti_6Si_7$, etc.

Results of investigation of physical-chemical properties, proceeding in the CP particles under conditions of plasma spraying, were used for optimization of technological parameters in production of coatings.

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DIFFUSION WELDING OF METALS AND ALLOYS

Technological processes of diffusion vacuum welding of different combinations of metals and alloys (copper-tungsten, copper-stainless steel, aluminium-titanium, bronze-steel, heatresistant alloys of different systems of alloying and others) with and without use of interlayers have been developed.

Technological processes and designed equipment guarantee the required properties and high quality of the joints.

Purpose and application. Technological processes are designed for welding special products, including those from dissimilar metals and alloys. They are used in different branches of machine building, for example, in manufacture of powerful cathodes of metallurgical plasmatrons, plate heat exchangers of high pressure, parts of precision pairs of friction of hydraulic units, parts of a hot path of gas turbines and so on.

Status and level of development. Technology and equipment for manufacture of plate heat exchangers are implemented in OJSC «Kriogenmash», bronze-steel parts of hydraulic units at Kharkov machine-building plant «FED».

Proposals for co-operation. Signing of contract is possible.

Main developers and performers: Prof. Yushchenko K.A., Dr. Nesmikh V.S., Lead. Eng. Kushnaryova T.N.



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TITANIUM. PROBLEMS OF PRODUCTION. PROSPECTS. Analytical Review. Part 2^{*}

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Methods of production of titanium ingots with application of both consumable and non-consumable electrodes are considered. Main problems, occurring in production of titanium ingots, are shown. Advantages and shortcomings of different methods are noted.

Keywords: titanium, melting methods, consumable electrode, non-consumable electrode, scrap, problems of production

Numerous methods of production of titanium ingots and alloys may be conditionally divided into two big groups ---- the group, in which a consumable titanium electrode is used, and the group, in which methods of direct remelting of the sponge and scrap are used. Different versions of such technologies (vacuum-arc, electron beam, electroslag, plasma-arc, and induction remelting) are described in detail in literature [1--4], that's why we will note only main issues and difficulties of the titanium melting connected with its properties, and quality requirements established for the ingots.

Titanium relates to chemically active metals, that's why its remelting, melting, and pouring are performed in vacuum or shielding environment. At high temperatures, not mentioning the molten state, titanium intensively reacts with all refractory materials including oxides, which usually enter into composition of the refractory materials, due to which it gets contaminated by oxygen. Melting of titanium in graphite crucibles causes its contamination by carbon. Content of impurities in titanium significantly influences its mechanical properties. There are interstitial (oxygen, nitrogen, carbon, hydrogen) and substitutional (iron and silicon) impurities. Impurities increase strength of titanium, but at the same time reduce its plasticity, whereby more negative effect exert interstitial impurities, especially gases. In case of introduction of just 0.003 % H, 0.02 % N or 0.7 % O, titanium completely loses its capacity for plastic strain and undergoes brittle failure [5]. Especially harmful is hydrogen, which reduces impact toughness and causes hydrogen brittleness of titanium alloys. That's why content of impurities, first of all gases, in titanium and its alloys (Tables 1, as well as Tables 11 in Part 1 of the Review) is strictly limited.

The need in using expensive equipment for manufacturing of the electrodes is one of the main shortcomings of mentioned technology. In addition, because of a limited possibility of the titanium waste introduction into the electrode-billet charge, significant consumption of primary titanium sponge and high cost of the item are characteristic of this technology. Inclusion into the charge of more than 30--35 % of waste reduces strength of the consumable electrodes and creates hazard of their failure during transportation and in melting [2]. To increase strength of the electrodes, 0.3--0.6 % Al is added into the charge. Requirements, established for the titanium sponge quality concerning, first of all, chlorine content, are connected with such phenomenon as ionization of a discharge, which can cause in VAR occurrence of the arc discharge between a consumable elec-

Most actively gases (especially hydrogen) and moisture, as well as residues of magnesium and salts, are removed from the charge materials in vacuum. That's why for production of the titanium ingots the technology of vacuum-arc remelting (VAR) is most widely used. At present at Verkhnyaya Salda Metallurgical Production Association (VSMPA) ingots of 450--1150 mm diameter of up to 9.5 t mass are melted, and melting of the ingots up to 18 t mass is possible by means of, as a rule, double or triple remelting of the consumable electrode. The task of the first remelting consists, as a rule, in removal of the reducer-metal, its chloric salt, hydrogen, and comparatively uniform distribution of the alloying additives. Quality of the ingot surface of first remelting does not matter much [1]. The goal of subsequent remelting operations consists in increasing uniformity of distribution of alloying additives and formation of ingots with homogeneous structure and good surface. As a consumable electrode for first remelting the ingot, pressed from the sponge, titanium waste and required amount of the alloying components, is used. A billet of 350--650 mm diameter and length up to 5500 mm is produced on a specialized hydraulic presses, which develops force up to 10000 tf [6].

^{*}Part 1 see in No. 1, 2007.

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Alloy grade	Al	V	Mo	Zr	Mn	Cr	Si	Fe	0	Н	Ν	С
VT1-00							0.08	0.15	0.10	0.008	0.04	0.05
VT1-0							0.10	0.25	0.20	0.010	0.04	0.07
ÂÒ1-2							0.15	1.50	0.30	0.010	0.15	0.10
OT4-0	0.41.4			0.3	0.51.3		0.12	0.30	0.15	0.012	0.05	0.10
OT4-1	1.52.5			0.3	0.7-2.0		0.12	0.30	0.15	0.012	0.05	0.10
OT4	3.55.0			0.3	0.82.0		0.12	0.30	0.15	0.012	0.05	0.10
VT5	4.56.2	1.2	0.8	0.3			0.12	0.30	0.20	0.015	0.05	0.10
VT5-1	4.36.0	1.0		0.3			0.12	0.30	0.15	0.015	0.05	0.10
VT6	5.36.8	3.55.3		0.3			0.10	0.60	0.20	0.015	0.05	0.10
VT6s	5.36.5	3.54.5		0.3			0.15	0.25	0.15	0.015	0.04	0.10
VT3-1	5.57.0		2.03.0	0.5		0.82.0	0.150.40	0.200.70	0.15	0.015	0.05	0.10
VT8	5.87.0		2.83.8	0.5			0.200.40	0.30	0.15	0.015	0.05	0.10
VT9	5.87.0		2.83.8	12			0.200.35	0.25	0.15	0.015	0.05	0.10
VT14	3.56.3	0.91.9	2.53.8	0.3			0.15	0.25	0.15	0.015	0.05	0.10
VT20	5.57.0	0.82.5	0.52.0	1.5-2.5			0.15	0.25	0.15	0.015	0.05	0.10
VT22	4.45.7	4.05.5	4.05.5	0.3		0.5-1.5	0.15	0.501.50	0.18	0.015	0.05	0.10
PT-7M	1.82.5			2.03.0			0.12	0.25	0.15	0.006	0.04	0.10
PT-3V	3.55.0	1.22.5		0.3			0.12	0.25	0.15	0.006	0.04	0.10
AT3	2.03.5					0.20.5	0.200.40	0.200.50	0.15	0.008	0.05	0.10
<i>Notes.</i> 1. B alloy 0.1	ase titaı wt.%.	nium; VT5-1	alloy inclu	des 23 w	t.% Sn. 2.	Total of a	ll other impu	rities in all a	lloys co	nstitutes	0.3, in	VT1-00

trode and the mould wall, and under certain conditions ---- an accident [7].

In case of using the technology of the consumable titanium electrode VAR in a water-cooled mould, one of the main shortcomings is a relatively small pool of molten metal, which complicates production of the ingots having homogeneous chemical composition and structure. We should note a dual approach to estimation of the metal pool size from the viewpoint of the metal quality. On one hand, for producing a homogeneous composition in remelting of a consumable electrode pressed from the titanium sponge, a deep pool and, therefore, a higher degree of the composition averaging is preferable because of inhomogeneity of the sponge and presence in such electrode of the wastes and alloying elements. On the other hand, for formation of a fine crystalline structure a shallow pool is more advisable. That's why in first VAR a big depth of the pool is ensured for achieving maximal possible chemical homogeneity, and in second VAR depth of the pool is reduced for producing a finer structure of the ingot. Unfortunately, in VAR a rigid relation exists between rates of the electrode melting and the ingot solidification, that's why possibility of exerting influence on the solidification process is very limited. In the process of remelting only arc current is regulated, while depth of the pool in VAR is directly proportional to the arc current value.

In VAR direct current sources are used, because the direct current arc in vacuum is stable, and application of the ingot as the anode allows increasing the metal pool depth due to efficient use of the energy, produced by the arc (on the anode, approximately, 75 %). The most essential shortcoming of using direct current sources in VAR is interaction of electromagnetic field of the direct current arc with the current fields induced in the metal masses located near the melting zone. Because of such interaction, throw-over of the arc on the mould wall and its burn-through is possible. For stabilization the arc is subjected to action of the longitudinal magnetic field created by different solenoids (both mobile and stationary ones, wound on the mould).

It is possible to remove an unbalanced magnetic field in the melting zone and exclude hazard of the mould burn-through in case of operation on the threephase alternative current. However, simultaneous presence of three electrodes reduces coefficient of the mould filling, and application of the sector electrodes noticeably complicates process of their manufacturing and increases their cost. In addition, a certain difficulty exists in maintenance of the stable melting rate of the electrodes [4].

In [8] principal possibility of maintaining stable burning of the alternative current arc in vacuum in remelting of one titanium electrode without application of the arc stabilizers (oscillators and easily ionized salts) by use of the controllable parametric power sources is shown.

Parametric power sources for VAR were installed and commissioned as long as in 1964. However, operation on alternative current requires for a higher



voltage than on the direct current, which increases hazard of occurrence of the side arcs. The ingots, melted on alternative current, have a significantly worse surface. As it was noted in [9], an attempt to achieve a sufficiently stable arc with such source failed, that's why these works were discontinued.

Interaction of magnetic fields in VAR causes intensive rotation of the metal pool. Because of rotation and sputtering of the molten metal by the arc and evaporation of the melt components on the mould walls above the melt level, the skull-crust forms. External in relation to the ingot layer represents a condensate, composition of which includes all evaporating from the remelted metal elements, whereby their concentration in this place is significantly higher than in the very ingot. Because of a characteristic appearance skull in the head part of the ingot got the name «crown». When the metal level rises, the skull partially dissolves in it and fuses with the ingot. Loose conglomerate of the condensate and metal sputter significantly worsen quality of the VAR ingot surface. Increased contamination of the metal surface layer by non-metal inclusions, excessive enrichment with alloying elements, impurities and gases, and presence of the sub-crust defects stipulate the need in roughing of the VAR ingots at the depth up to 25 mm [4].

The VAR technology does not include possibility of forming in the ingots of inclusions due to not completely melted refractory alloying elements. In addition, in alloyed titanium alloys a certain limit mass of the VAR ingots evidently exists, above which increase of the segregation degree of the alloying components and impurities and a courser structure of the big ingot metal will negatively effect quality of the items [2].

For the purpose of overcoming technological shortcomings, peculiar to the vacuum-arc method of the titanium ingot production, it was suggested almost 50 years ago in the E.O. Paton Electric Welding Institute to use the electroslag process [10, 11]. Investigations of this process concerning melting of titanium showed that in case of use of the open ESR furnaces, slag protection alone is not sufficient. When the slag pool surface is open, absorption of the gases (oxygen, nitrogen, hydrogen) by the slag and saturation with them of the remelted metal takes place in case of contact of the molten slag with air. Increase of the slag pool depth did not bring positive results. The need occurred to isolate the slag pool from air.

Shielding of the slag pool by an inert gas (argon, helium) allows reducing degree of the titanium contamination by gases, but does not ensure the required low level of their content. That's why first titanium was not melted in open moulds with blowing of the shielding gas into the melting zone, but in air-tight electric furnaces with a stagnant inert atmosphere, for creation of which the furnace chamber was preliminary vacuumized and then filled with argon. This process was called the chamber ESR. Argon pressure in the furnace had to be not less than 50–80 mm Hg, because at this pressure and temperature of electroslag process of the titanium remelting the slag popples. Optimal pressure of argon in the furnace is 300--600 mm Hg. Design of the chamber ESR furnaces little differs from that of the VAR furnaces [12, 13]. Because of complexity of the equipment and a number of other reasons development of the chamber type ESR furnaces was suspended [14, 15]. Only recently started to appear new chamber ESR furnaces [16] and ESR process with application of the chamber type furnaces, for example, magnet-controlled electroslag melting of titanium [17, 18].

Despite certain possible saturation of titanium by gases, in addition to the chamber ESR also ESR in the open type furnaces was developed. In [13, 19] principle possibility of producing titanium ingots by ESR method in an open mould with shielding of the melting zone by argon was shown. However, as weight of the ingots increases, consumption of argon excessively grows, and difficulty occurs with prevention of suction of gases from the ambient atmosphere [15].

To exclude this phenomenon chamber-free ESR furnaces with a flux gate were used, the design of which allowed creating in the process of melting in the melting space a small excessive pressure of gases, which additionally prevented getting of nitrogen and oxygen from the atmospheric air into cavity of the mould. The flux gate, making melting space air tight, shields the slag pool and a heated up to high temperatures surface of the consumable electrode above it against saturation with gases. It should be noted that in remelting in the open type furnaces in certain cases for preventing saturation with gases the surface of a consumable electrode is covered by a special slush and fluxes. In case of application of the slag gate there is no need in such operation. Estimation of efficiency of application of the flux gate in titanium ESR, performed in [20], showed possibility of producing titanium ingots of high quality in conventional chamber-free furnaces.

One of the main elements of the electroslag technology, playing energy-transformation and heattransfer role, is slag. In [11] the conclusion was drawn that slag for the electroslag melting of titanium ingots, as well as for welding of titanium, should be oxygen-free. As far as at the electroslag process temperatures titanium should not noticeably react with (calcium, strontium, boron) fluorides, it was suggested to use pure fluoric calcium, which got name ANT-2. Later composition of the slag was corrected (mainly for the purpose of the electric conductivity reduction), different additives were introduced, for example, CaCl₂, BaCl₂ [21], SrCl₂ and MgTiF₆ [17, 21], but the CaF_2 base was constant. As it was noted in [1, 22], electric conductivity of slag with 87.3 % CaF₂ at 1700 °C equaled 5.5 Ohm⁻¹ cm⁻¹. As amount of CaF_2 in the slag increases, its electric conductivity gets higher and, therefore, increases energy consumption in the melting.

Peculiarity of the ESR consists in absence of vacuum and need in using consumable electrodes with low content of hydrogen, because its weight share as a result of ESR, in contrast to VAR, practically does not reduce. Content of hydrogen in the sponge at once after vacuum separation usually does not exceed thousandth shares of percent [11, 23]. However, in the process of its crushing, sorting, storage, and fabrication of consumable electrodes saturation of the sponge with moisture and, in addition, absorption of moisture from the air by the ready, pressed from the sponge electrodes, are possible. Absence of the titanium degassing in the course of remelting is the main shortcoming of the electroslag process. That's why ESR of consumable electrodes is used, as a rule, for producing ingots of the second remelting in combination with one of the methods of the first remelting in vacuum, ensuring sufficiently full removal of gases and, first of all, hydrogen [1, 2].

Advantages of the consumable electrode ESR in comparison with VAR consist in the following:

• possibility of not using direct current sources;

• absence of arc in the process, which allows avoiding sputtering of the metal and occurrence of the ingot «crown» saturated with impurities;

• possibility of performing the process at a deeper molten metal pool and thus improving chemical homogeneity of the ingot, and getting due to more uniform heating over cross-section of the ingots of a rectangular section, which is convenient for their further rolling;

• possibility in a number of cases of avoiding roughing of the ingot surface before their subsequent processing.

In addition, recent investigations of the titanium consumable electrode ESR showed that there is possibility of dissolution and/or even refining and uniform distribution of high-nitrogen inclusions over volume of the ingot [24-26]. Rate of dissolution of high-nitrogen inclusions in ESR can be under certain conditions almost twice higher than in VAR.

A peculiar method of the consumable electrode remelting (arc-slag remelting ---- ASR), which is a combination of the arc and the slag processes of remelting, is suggested in [27]. Its predecessor, evidently, should be considered the method, according to which experimental melting operations of titanium by electric arc under a layer of the slag were carried out [1, 28]. As far as melting was performed at 60--70 V voltage in argon at the pressure 360--500 mm Hg, one should assume that these were melting operations in the chamber ESR furnace. Significant increase of the melting rate at increased power of the arc, formation of the ingot with a rather even surface without a «crown», and contamination of all units of the furnace, including the vacuum ones, by a dust-like flux were noted.

In the suggested ASR method the arc burns between the consumable electrode end and the slag pool surface; the film and the molten metal drops, formed on the electrode end, interact with the gas atmosphere. In this case requirements to air-tightness of the melting space of the flux gate design (such gates appeared for the first time after development of the ASR method) are significantly higher than in ESR. Applicability of the ASR method for remelting titanium and its alloys was estimated in [29--31]. The greatest advantages of this method consist in higher than in canonical ESR quality of the remelted metal, 2--3 times lower specific consumption of slag, and 1.5 times lower consumption of electric power [31].

Different methods of the titanium melting with a non-consumable electrode allow avoiding problems, connected with fabrication and remelting of a consumable electrode, and involving into processing a greater amount of the titanium waste. These can be vacuum-arc, plasma-arc and electroslag technologies.

Tungsten and graphite, as well as copper watercooled electrodes, are used as non-consumable electrodes in remelting of titanium. However, in this case problems occur as well. So called non-consumable electrode is anyway subjected to erosion, due to which contamination of the titanium ingot by inclusions of the electrode material occurs. The tungsten electrode cracks and is scoured by sputters of the remelted metal, due to which tungsten microparticles may be found in the produced ingot. For reliable operation the material, from which a non-consumable electrode is manufactured, should have good emission properties and high heat resistance.

For improvement of the operation conditions thorium, lanthanum, or yttrium are introduced into the tungsten electrode. Such additives not just reduce the electron work function and ensure higher allowable density of the arc current, but also increase temperature of the tungsten recrystallization, i.e. prevent embrittlement and improve its stability.

In case of application of the graphite non-consumable electrodes, carbonization of titanium was detected, for reduction of which the graphite electrode is impregnated by titanium.

The most pure from the viewpoint of contamination of the metal being remelted are copper electrodes. Copper, as a more low-melting metal, in case of its getting into the titanium being remelted completely dissolves in it, forming with it solid substitution solutions. It is not as hazardous as tungsten. The main task in case of using copper non-consumable electrodes consists in bringing their erosion to the minimal allowable level, whereby having ensured required productivity of the process and service life of the electrode (plasmatron).

In VAR and PAR with a copper non-consumable electrode the measures are envisaged, enabling distribution of the arc heat flux, getting on the electrode, over maximal possible area of its working surface and/or reducing current density in the arc reference spot on the electrode [32]. So, in the installations of the Schlinger Company a copper non-consumable electrode of the «rototrod» type, installed at an angle to



the metal pool surface, is rotated around its own axis and in this way the arc reference spot is moved over working surface of the electrode.

In installations of the Westinghouse Electric Company the electrode of «durarc» type is used; rotation of the arc over circumferential surface of the electrode tip is performed by means of the magnetic field. At working currents up to 2 kA service life of mentioned electrodes did not exceed 40–50 h.

In arc heating sources with a plasma cathode [33, 34] uniform dispersion of active spots on the working surface of a non-consumable electrode and reduction of the released on it energy are achieved due to auxiliary low-ampere arc, which ensures required number of the charged particles in the near-electrode area for operation of the main arc. At currents up to 10 kA service life of such electrode achieved as much as 20 h.

Further improvement of the non-consumable electrode designs caused development by the Westinghouse Plasma Company of the plasma electrode (plasmatron), in which movement of the arc reference spot was performed in two directions. Intensive rotation over internal surface of the water-cooled tubular electrode was ensured by the magnetic field, and movement along working surface of the electrode ---- by a powerful gas flow, whereby the arc periodically interrupted and formed again in a narrow gap between two tubular electrodes. The process was dynamically repeated, which allowed avoiding localization of the spot on the electrode surface. Service life of the plasma non-consumable electrode of such design of up to 4 MW power achieved several thousand hours.

To reduce current density in the arc reference spot it is necessary to ensure maximal possible diameter of the active spot, and best of all ---- the diffusion tie-up, or divide the spot into several spots uniformly dispersed over working surface of the electrode. Dispersion of the spot over working surface of the electrode enabled creation of a multi-rod or sectional electrode, consisting of a plurality of electrodes of small diameter gathered into the package [35]. The nearcathode discharge area of such electrode is formed from separate discharges (on each electrode of small diameter) united at a certain distance from the sectional electrode surface. Service life of such sectional electrode, in comparison with a usual massive one, at the same current is noticeably higher.

As far as melting technology of titanium electrodes with application of non-consumable electrodes is concerned, the following should be noted. Arc remelting in argon or helium and titanium VAR using non-consumable electrode in a copper water-cooled mould were used at early stages of development of the technology for production of titanium ingots. Because of contamination of the ingot by inclusions of the electrode material, a shallow metal pool and complexity of introduction of alloying components in the needed ratio, formation of inhomogeneity both over section and height of the ingot, poor penetration and, therefore, poor surface of the ingot (in roughing up to 25 % of metal turned into chips [1]), inefficient use of the released arc energy (approximately one third of it went on heating of the non-consumable electrode, and as residual pressure reduced this amount increased), and slow melting of the charge, these methods were used for a short time.

Due to improvement of the arc melting technology the described above consumable electrode VAR method and the vacuum-arc skull melting started to be used. In VAR a big pool of molten metal can be arranged in the skull, which makes it possible to use this method for production of not just ingots, but also of shaped castings. Peculiarities of thermo-physical processes in melting in skull are such that accumulation of the necessary amount of molten metal has to be ensured without its noticeable transition into solid phase (skull), i.e. with minimal losses. To achieve this it is necessary to wage the process at a higher rate of melting at high specific power of the arc. Skull vacuum-arc furnaces can operate both with non-consumable and consumable electrodes and combinations thereof. Current density in the skull furnaces is 1.5--2 times higher than in the furnaces for melting of ingots. Volume of molten metal in the crucible continuously increases, that's why temperature field of the pool in melting of the consumable electrode is always nonstationary, and area of the phase interface also continuously increases. Only in melting with a non-consumable electrode stationary heat state can be achieved [9].

In the skull furnaces copper and graphite crucibles installed in a water-cooled housing are, as a rule, used for melting of titanium. In melting in the metal crucible thickness of skull can be rather small. In case of full penetration of the skull, on the well cooled wall immediately forms a new layer of skull. However, in such crucible heat losses are higher than in the graphite one.

In melting in the graphite crucible skull thickness should be sufficiently big to prevent carbonization of the melt by carbon diffusion through the skull. In practice skull thickness in melting in the graphite crucible should constitute not less than 12--15 mm. Melting of new metal is usually performed in the available skull, which is continuously warmed up in the process of melting. At the beginning of melting thickness of the skull can somewhat increase and then (by means of warming up) reduce. The melting is performed in such way that thickness of the skull not to be changed from a melting to a melting.

Conditions of the skull cooling on the bottom and side walls of the crucible are different. On the bottom surface the skull is usually well fused with the crucible, while on the side surface it is fused to lesser degree. Big amount of fine pores forms in the places of the skull contact with the crucible. Because of presence of the pores and poor cooling the skull can be penetrated and the melt will get in contact with the crucible. Size of the crucible is determined taking into account thickness of the skull on its bottom and



side walls and power of the installation. For low- and medium-capacity furnaces (up to 250 kg) ratio of the crucible diameter to its height is selected equal three, and for high-capacity furnaces ---- two. In case of correct choice of the ratio between dimensions of the crucible and power of the furnace, molten metal may constitute 80 % and more of its general amount in the furnace. The melt is usually agitated in the crucible for averaging its chemical composition and temperature. Even in absence of a solenoid a noticeable movement of the melt was noted in the skull furnace. The melt is intensively mixed both in horizontal and vertical planes.

Skull melting is one of the methods, in which processes of melting and solidification of the metal are separated in time. That's why many defects of the ingot structure, connected with simultaneous melting and solidification of the metal in the same device, were not detected. For the ingots, produced by this method, conventional casting defects are characteristic aggravated by difficulties of the metal quality and temperature control.

As a result of the skull melting improvement, furnaces with an intermediate unit, or hearth skull furnaces, appeared. Melting in them is performed in a separate rather wide unit, the metal from which is poured into a non-cooled mould. Melting in the hearth skull furnaces was called the GRE process.

Production of titanium ingots using GRE process is one of technological schemes of melting-casting production of VSMPA. According to this scheme ingots are produced for their further remelting by VAR method along with traditional remelting of a consumable electrode pressed from the titanium sponge [36]. Application of the GRE furnaces allowed VSMPA involving into the turnover large amount of the own production wastes formed in manufacturing of items from titanium.

Plasma-arc technology of the titanium ingot production is characterized, in contrast to VAR, by a noticeable advantage ---- availability of the independent heat source, i.e. plasmatron, which allows regulating duration of the metal stay in molten state and depth and shape of the metal pool, and, therefore, targeted acting on the course of metallurgical processes and solidification of the metal.

Similar to the VAR process, in PAR direct current power sources are, as a rule, used. This is connected with the fact that the highest stability of the nonconsumable electrode (plasmatron) and stability of the plasma arc parameters are noted when the plasmatron is supplied with direct current of straight polarity. Application of direct current for power supply of non-consumable electrodes is inexpedient in development of big plasma furnaces, designed for melting of ingots, because of negative influence of the ferromagnetic masses located near short net of the furnace, and interaction of the magnetic fields of several simultaneously burning plasma arcs [4]. Taking into account stability of the non-consumable electrode, power of the melting furnaces is increased not due to increase of the plasmatron unit power, but due to increase of the number of plasmatrons. Application of several plasmatrons, possibility of separate change of their power and, therefore, rate of the metal melting provide the PAR process with a high technological flexibility. However, presence of several simultaneously burning direct current plasma arcs causes, because of interaction of their magnetic fields, sharp worsening of stability of the process parameters, change of the assigned position of the arc in the melting space, and intensive rotation of the metal pool. That's why in plasma complexes of several megawatts power alternative current sources with a three-phase power supply circuit are used [37]. In case of application of the alternative current sources, main difficulties cause change of burning conditions of the arc, accompanied by regular change of its polarity, and increase of the non-consumable electrode heat load at those half-periods, when the anode potential is fed to it. One manages to ensure stability of the alternative current plasma arc due to application of the power supply sources with significant reserve of the opencircuit voltage relative arc working voltage.

Technological scheme of PAR envisages remelting of both the consumable billet and the incompact charge. PAR proper of the consumable billet is application along with non-consumable electrodes (plasmatrons) of the consumable electrode, in preparation of which mentioned above problems occur. In addition, it is necessary to use in PAR (especially in the single-plasmatron furnaces with horizontal feeding of a billet) comparatively thin and because of this reason more labor-consuming in manufacturing and expensive consumable billets. As it was noted in [4], it is difficult for the PAR method to compete in remelting of a consumable billet with the VAR and the ESR methods. Energetic efficiency of PAR is lower, equipment for it is more complex, and it is necessary to use expensive inert gases. From economic viewpoint it is more expedient to use the PAR method for melting ingots from the charge materials. In relation to melting of ingots from titanium such technology was implemented as long as 30 years ago [38--40]. Equipment and technological process for PAR of the charge, consisting by 100 % of conditioned wastes, were developed. As showed numerous investigations [38, 39, 41, 42], titanium sponge of fine fractions, cast titanium waste, and wastes in the form of chips and cuttings can be remelted by the PAR method into ingots of primary and secondary titanium and its alloys.

As in any PAR process, problems, connected with application of the plasma gases, use of which not just complicates the PAR equipment, but also increases its remelting cost, remain unsolved. However, it should be remembered that an important aspect in production and application of titanium is reduction of the cost of its process stages.

In low power plasma-arc furnaces consumption of gas is not high, and usually it is not utilized after it



Melting conditions	Consumable electrode	Non-melting electrode
Current, A	25002700	1600
Voltage, V	23-25	2426
Electrode diameter, mm	60	30
Slag pool depth, mm	25-30	2530
Rate, kg∕ h	40.5	3.6
Specific consumption of electric power, $(kW{\cdot}h)/kg$	1.62	11.2

 Table 2. Technical-economic parameters of ESR of titanium ingots of 90 mm diameter by consumable and non-melting electrodes

 [11]

escapes from the chamber. As power and number of the plasmatrons in the furnace increase, gas consumption gets higher, and its discharge into the atmosphere is unprofitable. For repeated use of the gas, plasma-arc furnaces are equipped with the recirculation systems, which envisage gas scrubbing from solid and gaseous impurities, contaminating it in the process of the metal remelting. The lower is content of undesirable impurities in the plasma gas, the higher can be purity of the remelted metal.

In remelting of titanium, argon, initial degree of contamination of which may exceed allowable limits, is, as a rule, used as the plasma gas [23]. It especially concerns content of water vapor in it. In addition, in the course of its preparation for melting (pumping out of air from the furnace chamber by the roughing-down pump and its double flushing by argon) in the melting space remain about 0.1 % O and 0.5 % N.

In the process of titanium remelting, especially of the sponge, atmosphere of the furnace is saturated with removed from it gaseous impurities and salts. To maintain inert atmosphere, in the furnace chamber conditions of so called dynamic chemical vacuum are created. «Dirty» argon is removed into the regeneration system, and its space is occupied by the new scrubbed argon. Most frequently PAR is performed at atmospheric or a little higher pressure to prevent air suction into the furnace. From the viewpoint of interaction of undesirable gaseous impurities present in the furnace chamber in the titanium melt, it is advisable to maintain not atmospheric or excessive pressure in the furnace chamber, but a certain vacuum (about 300--400 mm Hg). However, it is possible to create such vacuum only in low-power plasma-arc furnaces, which discharge used gas into the atmosphere. On big PAR furnaces it is necessary to maintain excessive pressure in the furnace chamber for operation of the plasma gas recirculation system. If to compare metal purity, the metal remelted in argon is more contaminated with nitrogen and non-metal inclusions than in case of VAR [4].

ESR with application of the non-consumable electrode is used, as a rule, for remelting of the charge materials and utilization of chips and small cuttings. This process is also used for replenishment, heating and melting of ingots, and refining of the metal [43, 44].

As long as 50 years ago in the process of investigation of the electroslag process applicability for production of titanium ingots, efficiency of the titanium ESR with a consumable and non-melting (non-consumable) electrode was compared [11]. Results of such comparison are presented in Table 2.

In melting of small (90 mm in diameter) titanium ingots with application of the metal water-cooled nonconsumable electrode of 30 mm diameter consumption of electric power turned out to be unreasonably high, while productivity was insufficient. Also insufficient stability of the electrode was noted. Unfortunately the authors did not mention the material, from which the non-consumable electrode was manufactured, but it was, evidently, a copper water-cooled electrode. One can see from Table 2 that melting current is 1.6 times lower in case of using a non-consumable electrode, than in case of using titanium electrode of a bigger (60 mm) diameter.

Application of a non-consumable electrode of smaller diameter is, in all probability, connected with convenience of feeding into the gap between the electrode and the mould wall of the charge materials, but such statement is not quite correct. Reduction of the non-consumable electrode diameter caused significant increase of the current density in it. Processes of electrocorrosion and electroerosion of such electrode in the molten slag were, evidently, so intensive, that increase of the current above mentioned level was found unadvisable. A lower value of current and presence of a water-cooled electrode in center of the slag pool, causing additional heat losses, stipulated reduction of melting rate of the metal being remelted and, therefore, increase of specific consumption of electric power. Application of non-consumable electrodes in the classic ESR system in general and titanium ESR in particular, were considered at that time irrational [1].

However, development of new designs of non-consumable electrodes [45] and later experience of application of the metal water-cooled electrodes in electroslag processes [46--48] showed that this method, not being inferior to conventional ESR with application of a consumable electrode by quality of the produced metal, exceeded it by the technological possibilities [47]. The most rational design of a watercooled non-consumable electrode is considered the design with replaceable tips. Such tips, to copper body of which over their end and side surfaces molybdenum and tungsten rods are brazed (the design was, evidently, taken from the PAR experience), ensure service life at maximal loads up to 300 h. The electrodes are simple in servicing, reliable in operation, and ensure stable performance of the electroslag process.

It should be noted that electroslag process with application of the metal non-consumable electrodes is performed in such way that a layer of the slag skull, «frozen» on the water-cooled electrode tip, be constantly maintained. In case of non-observance of the energy conditions, the slag skull melts and the tip starts to work in contact with the molten slag, due to which its service life reduces and probability appears of the melt contamination by the tip material. Application of such non-consumable electrodes in melting of ferrotitanium is described in [49]. As it is noted in this article, content of titanium over diameter of the ingot depends to a great degree upon scheme of the current supply to the slag pool. In this case its more uniform distribution was achieved in the monofilar scheme of supply with a split electrode, in which uniform heat release in the slag pool and better conditions for melting of the metal and heat transfer to the metal pool were ensured. Depth measurements of the latter showed that in case of the monofilar scheme it is 1.2--1.5 times deeper than in the bifilar scheme, and its depth increases near walls of the mould, which enables averaging of the ferrotitanium composition in this zone.

The need of using slag with high electric resistance, when higher values of the slag temperature, productivity of remelting, and lower consumption of electric power are achieved, was noted. Commercial electroslag remelting process with application of nonconsumable water-cooled electrodes for production of ferrotitanium was implemented at VSMPA. At present ferrotitanium is produced at VSMPA by the electroslag method [36].

Remelting in the current-conducting mould should, probably, be also regarded as the ESR method with application of a non-consumable electrode [50, 51]. The upper current-conducting section of such mould is a non-consumable electrode, erosion of which, taking into account a significantly bigger area of the working surface, is significantly lower than in a conventional copper electrode of a relatively small diameter, located in the slag pool center. But in this case protection coatings of the current-conducting surface are used for protection against electroerosion and electrocorrosion as well [52].

Authors of one of the first designs of the currentconducting mould [51] suggested using tungsten, molybdenum or graphite as a protection coating. Development of different technological processes in the mould, especially processes with pouring of the molten metal [53], caused appearance of the bimetal structures with protection of a copper current-conducting surface by a relatively thin layer of the carbon steel [52, 54]. In case of melting of titanium and its alloys titanium may be used as a protection coating.

Application of the current-conducting mould for melting of titanium sponge is described in [55, 56].

The melting is performed in the chamber furnace in argon. It should be noted that in case of the sponge remelting in the current-conducting mould it is advisable to use slag with higher values of the melting point and electric resistance than in usually used for this purpose fluoric calcium. As diameter of the mould increases, problems may occur with melting of the sponge, non-melted parts of which accumulate in the forming part of the mould, because of insufficient temperature of the slag pool. For intensification of the sponge melting process it is possible to use a double-circuit scheme of remelting [57, 58], one of the versions of which represents a combination of the processes of remelting in the current-conducting mould and remelting with application of a non-consumable electrode, located in the slag pool center. Possibility of using such scheme for remelting of titanium sponge is shown in [56]. Availability of additional copper water-cooled electrode returns us to the problems, connected with its stability, heat efficiency of the process, and other above described problems. It is more advisable to use a consumable electrode in the double-circuit scheme.

Different methods of induction and electron beam remelting allow refusing in general from application of both consumable and non-consumable electrodes. These methods and prospects of the titanium industry development will be considered in the third part.

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MASTERING OF PRODUCTION TECHNOLOGY AND INVESTIGATION OF QUALITY OF ESPECIALLY LOW-CARBON ROLLED WIRE

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Technology of especially low-carbon steel production in the arc steel furnace is mastered. Parameters of electric furnace intermediate product, degassing of which allows producing especially low-carbon steel, are experimentally substantiated. The determining factor is achievement of the optimal oxidization of the tapped metal for removal of carbon. Complex of investigations of the macro- and microstructure, degree of inhomogeneity and contamination with non-metal inclusions, and mechanical tests of the rolled wire showed that the structure and level of the properties create possibility for its drawing without intermediate annealing.

Keywords: especially low-carbon steel, degassing, out-offurnace treatment, non-metal inclusions, mechanical properties

Moldavian Metallurgical Plant (MMZ) ---- one of the most advanced enterprises of CIS ---- constantly expands assortment of the produced rolled stock. Due to significant technological possibilities of production, high skill of specialists of the enterprise and attraction of scientific developments, the task of production from the electric furnace intermediate product of especially low-carbon steel for manufacturing highplasticity rolled wire was recently solved.

The technology for especially low-carbon steel production was developed with application of thermodynamic calculations on the basis of the active experiment, which included production of the low-carbon intermediate product in the arc steel furnace (ASF), its out-of-furnace treatment on the ladle--furnace installation (LFI) and vacuum apparatus of chamber type (VD), and a complex of investigations of the metal quality (mechanical properties, amount and morphology of the non-metal inclusions).

The main technological complications of such metal production are achievement of low (less than 0.3 %) weight share of carbon in the steel tapped from ASF, and in out-of-furnace treatment in the ladle with periclase-carbon lining, and production of the dense structure of the billet from the metal with low amounts of silicon and manganese (less than 0.02 and 0.12 % respectively) and limited consumption of aluminium.

The developed technological process for production of especially low-carbon steel included melting of the intermediate product in ASF from the metal scrap with addition of a portion of non-conversion iron (or some other primary raw material) for ensuring minimal content of non-ferrous metals (chromium, nickel, copper); tapping of the intermediate product into the ladle and its partial reduction; decarburization of the metal on the degassing installation of the chamber type; bringing of the metal up to the assigned chemical composition in the ladle--furnace installation; casting of steel on MCCB with full protection of the jet against secondary oxidation and application of the system for electromagnetic mixing of metal in the mould.

Melting of intermediate product in arc steel furnace. Arc steel furnace of MMZ is a state-of-the-art high-productivity unit, operating with intensification of the melting process by means of oxygen injection and feeding into the furnace of the carbon-containing materials (coke, natural gas) using several kinds of burners and lances. Oxygen, injected into ASF allows performing controlled decarburization of the melt, reburning CO into CO₂ in working space of the furnace, performing heating and cutting of the metal charge by means of different fuel burners, and improving slag foaming due to generation of the CO bubbles. After melting of the metal charge, when content of carbon in the melt is high, practically the whole oxygen goes for decarburization of the molten pool, and at the last stages of the oxidation refining (< 0.1 % C) preferential oxidation of iron occurs. In production of especially low (0.02--0.03 %) content of carbon up to 90 % of oxygen is consumed for formation of iron oxide, high content of which in the slag causes reduction of the lining stability.

For producing in ASF of the intermediate product fit for production of especially low-carbon steel, it is necessary to investigate possibility of producing minimal possible amount and substantiate maximal allowable content of carbon in the tapped metal. For this purpose a series of melting operations of the low-carbon steels was performed (Table 1).

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Consumption of oxygen and fuel							Parameters at tapping		
General O₂, m³∕t	Coke, kg∕t	O_2 for combustion of coke, m^3/t	C_3H_4 , m ³ /t	O_2 for combustion of CH ₄ , m ³ /t	O_2 for oxidation, m^3 / t	C, %	aî, ppm	<i>T</i> , °C	
51.0	8.21	7.66	7.3	15.64	27.69	0.029/0.025	1799	1741	
50.6	8.70	8.15	7.3	15.64	26.81	0.043/0.029	1470	1694	
53.6	13.57	12.67	7.3	15.64	24.39	0.051/0.025	1741	1707	
36.0	10.22	9.54	7.6	16.29	10.18	0.055/0.035	1145	1653	
43.7	6.50	6.07	7.8	16.71	20.92	0.096/0.032	1299	1701	
44.1	5.79	5.40	8.0	17.14	21.55	0.076/0.038	1118	1719	
56.9	24.51	22.88	7.2	15.43	18.60	0.074/0.033	1213	1644	
52.9	29.76	27.78	6.3	13.50	11.62	0.054/0.026	1445	1703	
54.9	17.42	16.26	6.2	13.29	25.36	0.038/0.030	1316	1660	
Note. In nume	erator data on o	chemical analysis, in	denominator o	lata of oxidation	sensor CELOX	are presented.			

Table 1. Consumption of oxygen and fuel in ASF and certain parameters of melts of experimental series at tapping

In all melts, despite high amount of carbon-containing materials injected into the furnace for heating the metal, rather low final content of carbon (0.030-0.096 % according to the chemical analysis and 0.025-0.038 % according to the CELOX sensor for measurement of the molten steel oxidation) and high values of oxidation and temperature (1118--1799 ppm and 1653--1741 °C respectively) were obtained.

For the investigated melts share of oxygen used for oxidation of impurities (difference between general amount of injected into the furnace oxygen, and that consumed for combustion of coke (C = 95 %) and of methane ---- according to stoichiometry) was calculated. It is shown that between consumption of oxygen above stoichiometry of the fuel combustion reactions and oxidation of the tapped metal (Figure 1) rather close relation existed.

The data obtained show possibility of the oxidation (carbon content) control at the outlet from ASF. Achievement of a significant degree of oxidation is undesirable from the viewpoint of reduction (melting loss of iron) of the efficient metal output and the lining stability. However, one has to take into account both initial content of carbon in the intermediate product and its further entry into the metal after its tapping from the furnace. So, for production of espe-



Figure 1. Dependence of oxidation *a*O of metal at tapping upon general consumption of oxygen (\blacklozenge) and its calculated amount (\blacktriangle) used for oxidation of impurities

cially low-carbon steel, oxidation of the intermediate product should ensure removal of carbon from the metal in vacuum treatment up to such level, at which its further entry (from ferroalloys during reduction of steel, from electrodes during heating of steel in the ladle--furnace, and from periclase-carbon lining of steel ladle) will not cause excess of the carbon content targeted values in the metal.

Calculation of minimally necessary content of active oxygen in the metal for getting 0.005 % C in steel at its different weight shares in the intermediate product was carried out (Figure 2).

Comparison of the values of necessary oxidation level according to the calculation (80--1000 ppm) with their actual range in the intermediate product at tapping from ASF (458--1997, average ---- 1015 ppm according to the statistical data analysis of 200 melts of the current production and results of experimental testing) shows that real oxidation level of the tapped metal is sufficient, sometimes even excessive from the viewpoint of carbon removal from the initial intermediate product in subsequent vacuum treatment of the metal. Content of oxygen in the steel, necessary for removal of carbon entering with ferroalloys (ferromanganese ---- 7, silicon manganese ---- 2 %) for ensuring required concentrations of manganese and





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Figure 3. Equilibrium content of oxygen O_{eq} at different values of temperature and weight share of carbon in metal

silicon in the steel, was calculated taking into account degree of assimilation of the leading elements. It is established that in case of use of both silico- and ferromanganese, active oxygen contained in tapped from ASF metal, is able to oxidize carbon brought into the metal with ferroalloys and lining of the steel ladle.

In out-of-furnace treatment of the produced intermediate product (degassing and finishing on LFI) final content of carbon in the metal less than 0.01 % is achieved, even when its initial weight share (according to the chemical analysis data) achieves 0.006 %, and oxidation level of the tapped from the furnace metal exceeds 1100 ppm. Peculiarities of the out-of-furnace treatment processes (degassing, desulfation, and reduction) for production of especially low-carbon steel were investigated on melts Nos. 1--3. Out-of-furnace treatment of the intermediate product was performed according to the reverse scheme ----ASF--VD--LFI--MCCB.

Degassing of intermediate product for production of especially low-carbon steel. Calculated substantiation of carbon content reduction in degassing was performed. Experiments on production of especially low-carbon steel by vacuum-oxygen decarburization of the initial intermediate product (without injection of oxygen) were carried out. It was taken into account that temperature of the metal tapped from ASF was significantly higher than 1600 °C, for which main thermodynamic parameters of the steel decarburization reaction, presented in majority of literature sources, were calculated. Equilibrium values of oxygen content in the metal at 0.0020--0.0016 % C for the temperature range 1550--1750 °C were established. Graphic presentation of the results obtained is given in Figure 3.

Under normal pressure conditions reduction capacity of carbon at the concentration above 0.08 % exceeds that of manganese, but is lower than that of silicon (at the same concentrations). Pressure reduction enhances reduction capacity of carbon, whereby significant influence is exerted by the temperature, increase of which shifts equilibrium of the vacuum-oxygen decarburization reaction. As a result reduction capacity of carbon gets higher than that of manganese and silicon at the same concentrations even at the temperatures above 1600 $^{\circ}$ C and pressure below 100 mbar.

It is established on the basis of calculation of the oxygen equilibrium concentrations within temperature range of the MMZ technological process for steel of the targeted chemical composition 0.01 % C, 0.12 % Mn, 0.02 % Si that pressure reduction in the vacuum chamber down to 100 mbar is sufficient for prevailing oxidation of carbon in comparison with manganese and silicon within the whole range of the technological process temperatures.

The metal, partially reduced by aluminium at tapping from ASF, was subjected to degassing. After installation of the ladle into the vacuum chamber, averaging blowing of the metal by argon was performed. Simultaneously the samples were taken, and temperature and oxidation of the metal were measured (Table 2).

In the cause of degassing of the melts no additives were introduced into the ladle. In the process of the vacuum development and during the whole period of treatment intensive «boiling» of the melt was registered, which is explained by the proceeding reaction of vacuum-oxygen decarburization of the metal. Rarefaction under deep vacuum conditions constituted 2.0--2.5 mbar. In further seasoning of the metal under vacuum the decarburization process gradually decelerated or stopped in general, which proved absence of one of the elements that participated in the reaction $C + 1/2 O_2 = CO$. As far as a limiting link in this

 Table 2. Main parameters of out-of-furnace treatment on VD

		Parameters of metal							Duration of dogassing		Argon consumption at	
Melt No.	Weight share of carbon, %		aÎ, ppm		Т, С		stages, min			stages, m ⁷ /h		temperature reduction in
	Before	After, less than	Before	After (act./calc.)	Before	After	All together	1	2	1	2	°C/min
1	0.0293	0.010	884	573/627	1721	1648	23	12	11	46	46	2.4
2	0.0327	0.010	644	472/341	1645	1576	25	10	15	2-3	46	2.3
3	0.0564	0.010	784	186/165	1676	1592	22	8	14	N/D	56	2.1
Average	0.0395	0.010	771	410/378	1681	1605	23	10	13	34	46	2.3
Note. At va	Note. At vacuum treatment stage 1 development of vacuum occurs, at stage 2 deep vacuum.											

reaction under our conditions is content of carbon, and its exact determination at low concentrations (less than 0.01 %) is difficult, change of the weight share of active oxygen was determined. Rather accurate ratio of the actual and calculated change of oxidation (see Table 2) and carbon content in the metal as a result of degassing in all melts was obtained. Calculated amount of CO released in vacuum treatment constituted in experimental melts from 36 to 86 m³, which significantly exceeded the volume of argon injected for mixing of the metal. So, surface of the gass-metal interaction is much more developed in degassing of the partially reduced metal, which is a prerequisite for removal from it of the impurity gases (nitrogen and hydrogen).

Technological peculiarities of ladle treatment of especially low-carbon steel. In addition to removal of carbon, the main task of treatment in the ladle-furnace in production of especially low-carbon steels is reduction of sulfur content in the metal down to the necessary level. Successful progress of the desulphurization process is ensured due to quick formation of the refining slag of the optimal chemical composition. To increases coefficient of sulfur distribution between the slag and the metal (sulfur-absorption capacity of the slag) it is necessary, first of all, to ensure high basicity of the slag and low content of iron oxide (low oxidation of the metal). Efficient desulphurization of the metal in its treatment under the basic slag is achieved, when content of oxygen in the metal does not exceed 100 ppm and that of iron oxide in the slag is less than 0.5 %. That's why for determining the conditions, under which intensive desulphurization of the metal is ensured, content of active oxygen in experimental melts in the course of out-of-furnace treatment was determined.

Peculiar feature of the molten especially low-carbon steel is limiting in the metal of not just carbon content (less than 0.01 %), but also of silicon (less than 0.02 %). Low concentration of the main deoxidizers requires for increased attention to oxidation of the metal. For final reduction of the metal in experimental melts calcium ferroalloys (silicocalcium and ferrocalcium) were used, which are characterized by high affinity to both oxygen and sulfur. So, minimization of their consumption is possible only at low contents of active oxygen and sulfur in the steel.

During development of the reduction scheme for especially low-carbon steel it was taken into account that reduction capacity of silicon at its weight share in the metal below 0.01 % can be implemented only provided oxidation of the metal is more than 295 ppm (Figure 4).

At initial concentration below 295 ppm oxygen practically does not react with silicon, that's why stronger reducers are required (aluminium, calcium). Application of 0.005 % Al ensures oxidation in the dissolved form at the level 11 ppm. In addition, one has to take into account that significant residual weight shares of aluminium in steel can be the reason



Figure 4. Results of calculation of equilibrium oxygen concentrations at characteristic for especially low-carbon steel concentrations of reducers *A*: *1* — manganese; *2* — carbon; *3* — silicon; *4* — aluminium; *5* — calcium

of pulling of the nozzle during casting of steel on MCCB. The lowest equilibrium content of oxygen (less than 1 ppm) can be ensured in case of application of calcium with its dissolution in the metal within the range of 0.0002--0.0010 %.

Example of oxygen balance, taking into account chronology of adding materials into the ladle, is given in Figure 5.

It is shown that all prerequisites for successful process of the metal desulphurization appear by the 10-th minute of treatment. Induced slag is characterized by high refining capacity, which ensures degree of desulphurization above 70 %, weight share of sulfur being less than 0.010 %.

Experimental testing confirmed negative influence of high content of active oxygen in the metal on desulphurization of the metal and irrationality of increasing basicity of the slag above 3.5 units. That's why it was recommended for the purpose of simultaneous reduction of these parameters to perform preliminary reduction of the metal by the normalized additions of ferrosilicon and aluminium and control observance of the oxygen and silicon balance before addition of the slag-forming components.

Casting of the melts was performed on MCCB through an intermediate ladle with a magnesian lining (diameter of the dosing unit channels is 16.5 mm) with full protection of the jets against secondary oxidation with application of the system for electromag-









Type of inclusions	Non-metal inclusions in melts according to ASTM E45, points					
	1	2	3			
Type A narrow (24 μm)	$\frac{0.5-1.0}{0.625}$	$\frac{0-0.5}{0.42}$	<u>00.5</u> 0.17			
wide (> 412 μ m)	0	0	0			
Type B narrow (2-9 μm)	$\frac{0.5-1.0}{0.375}$	$\frac{1.0}{1.0}$	$\frac{1.0}{1.0}$			
wide (> 915 $\mu m)$	0	$\frac{0-2.5}{0.83}$	$\frac{0.5-5.0}{1.75}$			
Type C narrow (2–5 μm)	$\frac{1.0-2.0}{1.375}$	$\frac{1.0-2.0}{1.25}$	<u>01.5</u> 0.92			
wide (> 512 $\mu m)$	$\frac{0-3.0}{1.75}$	$\frac{0-2.0}{0.33}$	0			
Type D narrow (2–8 μm)	$\frac{0.5}{0.5}$	$\frac{0.5-1.0}{0.75}$	$\frac{0.5-1.0}{0.83}$			
wide (> 813 $\mu m)$	0	<u>00.5</u> 0.33	$\frac{0-0.5}{0.25}$			

Table 3. Non-metal inclusions in metal of the especially low-
carbon steel melts according to ASTM E45

Note. In numerator variance of values and in denominator mean value are indicated.

netic mixing in the moulds. From all billets the rolled wire of 5.5 mm diameter was produced.

Results of investigation of the especially lowcarbon rolled wire quality. A complex of metallographic investigations of the billet and the rolled wire and mechanical tests of the rolled wire metal, produced in the test-commercial melts, were performed.

Microstructure of the metal in cross section of the rolled wire from especially low-carbon steel consists of ferrite and tertiary cementite, distributed relatively uniformly in the form of fragments of the lattice over grain boundaries. Axial segregation in all investigated sections of the rolled wire of experimental melts corresponded to the first class of the reference row according to EN 10016-1--1994. Size of the actual grain (in the rolled wire of 5.5 mm diameter according to GOST 5639--82 at magnification power 400) in the metal of all investigated bundles of melts 1--3 corresponded to 10--11 points. Average calculated mass of scale on especially low-carbon metal of the experimental melts equaled 13.8 kg/t.

Degree of contamination with non-metal inclusions according to GOST 1778--70 (method \emptyset) and

Table 4. Mean values of mechanical properties of rolled wire metal of 5.5 mm diameter from especially low-carbon steel

netar of 9.5 min diameter from especially fow carbon seen								
Melt No.	σ_t , MPa	σ _{0.2} , MPa	ψ, %	δ ₅ , %	δ ₁₀ , %			
1	354	239	85	43	34			
2	321	198	84	44	36			
3	362	245	84	42	33			

ASTM E45 (method A) was determined. Results of estimation according to ASTM E45 are given in Table 3.

It is shown that as to the content of non-metal inclusions especially low-carbon steel meets requirements of the standards established for steel of C4D type.

Microhardness (mean HV_{500}) in experimental melts 1--3 equaled 101.87, 93.22 and 108.89 respectively. Microhardness correlates well with the mechanical test results ---- tensile strength and yield, mean values of which are given in Table 4.

Produced rolled wire has low values of tensile strength, corresponding to the level of IF-steels, and high values of plastic properties, which makes it possible to perform drawing without intermediate annealing operations. Application of such high-plasticity rolled wire on metal-ware enterprises allows significant reducing level of breakage and makes it possible to significantly reduce expenses connected with intermediate annealing operations.

CONCLUSIONS

1. Parameters of the electric furnace intermediate product, fit for production of especially low-carbon steel after its subsequent degassing, are experimentally substantiated.

2. It is established that the most important is achievement of the optimal range of activity of oxygen in the metal at tapping, which ensures removal of carbon up to the assigned limits taking into account its additional inflow within the technological cycle.

3. Due to application of the vacuum-oxygen decarburization reaction (without additional introduction of oxygen in gaseous form or in the form of oxides) 0.01 % C content in the efficient metal is achieved, which at weight share of manganese up to 0.12 and silicon up to 0.02 % ensures favorable microstructure and high level of plastic properties of the rolled wire.

ON ISSUE OF DEPENDENCE OF THE CARBON GRANULAR MATERIAL ELECTRIC PROPERTIES UPON PRESSURE AND SIZE OF FRACTIONS

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Analytical model, describing dependence of specific electric resistance of the electro-conducting granular materials upon pressure, is obtained. It is shown that electric conductivity of the granular material is invariant in relation to the size of its fractions.

LECTROMETALLURGY

Keywords: granular material, contact spots, pressure, specific electric resistance

Electric properties of granular carbon materials have been intensively studied in Ukraine and abroad for a number of decades. Judging from the publications [1--12], a great number of investigations were carried out in this direction, and a significant experimental material was accumulated within recent years. However, there is no possibility of efficient use of the experimental investigation results because of absence of the appropriate mathematical models, which would describe in analytical form interrelations between electric, mechanical and thermophysical properties of the granular materials. In particular, when solving practical tasks, the need frequently occurs in finding out the issue, to what degree electric conductivity of the carbon materials depends upon degree of their crushing and external pressure on the crushed material. Presented in literature [1, 2, 6, 7] data on mentioned dependence are rather controversial.

In [8, 11] an attempt was made to build a mathematical model, which would reflect functional dependence of specific electric conductivity of thermoanthracite ρ_{gr} upon size of its fractions. In [11] a simple formula

$$\rho_{\rm gr} = \rho_0 \ln \frac{2r}{a} \tag{1}$$

was drawn, where ρ_0 is the specific electric resistance of the material itself; *r* is the grain radius of the crushed thermoanthracite; *a* is the reduced radius of so called contact spots (the Holm spots [13]).

Disadvantage of the formula (1) in its practical use is the fact that radius of the contact spots *a*, included into it, is not known in advance. Radius *a*, as it will be shown later, does not directly depend upon radius of the grains *r* and force F_c , with which the grains are pressed to each other, that's why the relation (1) actually does not give us the answer to the question as to real dependence of the specific

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electric resistance of the granular material upon the size of its grains.

In this work the task is put to find analytical relations, which would allow getting a more clear idea about the degree, to which specific electric resistance of a granular material depends upon external pressure and size of its fractions.

Let us estimate, first of all, radius of the contact spots a. The contact spots of grains of any granular electrical-conducting material are formed as a result of contact of the grains with each other under action of various forces applied to these grains. As it follows from the mathematic theory of elasticity [14], if two conglobate bodies of the same size, made from the same material, are pressed to each other with force $F_{\rm c}$, an elastic strain will occur in them, and after this the grains will touch each other not in a single point but over a certain small, but finite area of their surface. It is shown in [14, 15] that around the touch point the contact area, limited by a circumference, is formed (the Hertz problem). Radius of this circumference a, compression force F_c , and radius of the conglobate bodies r are interconnected by the dependence

$$a = \left[\frac{3F_{\rm c}r(1 - v^2)}{4E}\right]^{1/3},$$
 (2)

where v is the Poisson's ratio; E is the Young's modulus.

Let us use this result in the considered problem. Let us note that compression force F_c , included into the expression (2), is not known in advance and can not be directly measured. However, F_c can be expressed through the known to us external load F on the granular material, place into a certain vessel with a quite certain area of the cross-section S. If $S = I^2$, where I = 2rn, and n is the number of grains densely placed along a section of the line of I length, than it is evident that

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<i>F</i> , N	<i>P</i> , Pa	R ₁ , Ohm	R ₂ , Ohm	$\rho_{gr1} \cdot 10^{-3}$, Ohm·m	$\rho_{gr2} \cdot 10^{-3}$, Ohm·m
3.92	1265	2.15	2.04	78.5	74.4
9.80	3161	1.91	1.83	69.7	68.8
19.60	6322	1.63	1.71	59.5	62.4
29.40	9484	1.47	1.57	53.7	57.3
39.20	12645	1.39	1.45	50.7	52.9
45.08	14542	1.30	1.41	47.4	50.4

Results of experiment

$$F_{\rm c} = \frac{F}{n^2} = \frac{F}{l^2} 4r^2 = 4r^2 P \; .$$

Here $P = F/l^2$ is the pressure stipulated by external load F on unit of the area $S = l^2$.

Let us substitute into formula (2) instead of F_c equal to it value $4r^2P$ and write down expression for radius of the contact spots in the form of

$$a = r \left[\frac{3P(1 - v^2)}{E} \right]^{1/3}.$$
 (3)

One can see from formula (3) that radius of the contact spots a is directly proportional to radius of the grains r. It is an important issue.

Let us substitute the obtained expression (4) into formula (1). As the result we will get the required ratio

$$\rho_{\rm gr} = \frac{\rho_0}{3} \ln \frac{D}{P} \,, \tag{4}$$

where $D = 8E/3(1 - v^2)$ is the constant coefficient, which characterizes elastic properties of the material. It is not difficult to note that all values, which enter into formula (4), are, in contrast to the expression



(1), known in advance or can be rather easily measured.

So, it follows from formula (4) that specific electric resistance ρ_{gr} does not depend upon radius *r*, i.e. electric properties of the granular materials are invariant in relation to size of the fractions.

The established hear fact can be physically expressed in the following way. To the granular material grains of a smaller diameter r corresponds a greater number n thereof (because n = l/2r, and l = const). As the number of grains n increases, compression forces between the grains F_c (at F = const) reduce according to the relation $F_c = F/n^2$, and, therefore, also reduces radius of the contact spots a. This causes the situation, in which at any r values the relation r/a, according to expression (3), is the constant value, i.e. specific electric resistance of the granular material $\rho_{\rm gr}$ does no depend upon radius r of the grains.

This theoretical conclusion matches quite well experimental data presented in [1]. Results of the experiments, presented there, show that at change of the anthracite grain size within a rather wide range (from 0.015 to 20 mm), values of the specific electric resistance ρ_{gr} changed at a low external pressure P = 14.7 kPa only within the range (9.3–8.7)·10⁻³ Ohm·m, i.e. practically remained constant.

One can see from expression (4) that specific resistance ρ_{gr} depends upon external pressure *P*, but the very dependence is of logarithmic character. In construction of the mathematical dependences (1) and (4) the simplifying assumptions relative shape of the grains, their arrangement and contact surfaces were made. In this connection the question occurs to what degree the obtained expression (4) reflects real dependence $\rho_{gr} = \rho_{gr}(P)$. To give answer to this question the experimental check of this formula was carried out (4).

Like in [11], crushed thermoanthracite, produced at the enterprise «Ukrainian graphite», was used as the granular material. Mean values of the grain radii of two fraction groups, selected for the experiment, equaled $r_1 = 2.3$ mm and $r_2 = 9.0$ mm. In the process of the experiment external load F on the crushed thermoanthracite, poured into a special container, was changed; the area of cross-section and the height inside the container had the following values: $S = 3.1 \cdot 10^{-3}$ m², $l = 85 \cdot 10^{-3}$ m. Range of change of the load F was selected within 3.9--45.1 N in order deformation of



the grains, occurring in this case, to be significantly lower than elastic limit of thermoanthracite.

As a measurement value electric resistance of the crushed anthracite *R* was selected. Specific resistance $\rho_{\rm gr}$ and pressure *P* were calculated according to the formulas $\rho_{\rm gr} = RS/l$, $\rho = F/S$. Results of the experiments are presented in the Table.

One can see from the Table that values of specific resistances ρ_{gr1} and ρ_{gr2} practically do not differ from each other, while corresponding to them radii of the grains $r_1 = 2.3$ and $r_2 = 9.0$ mm differ significantly (almost 4 times). This fact, as well as experimental data of [1], is in full agreement with the statement about invariance of the granular material electric properties relative size of its fractions.

On the basis of the Table data points (designated by the circles) are drawn on the Figure, which characterize experimentally obtained dependence $\rho_{gr1} = \rho_{gr1}(P)$, and the solid line shows the diagram of dependence built according to formula (4) at $\rho_0 = 42.8 \cdot 10^{-3}$ Ohm·m and D = 0.38 MPa. Mentioned values of the parameters ρ_0 and D were determined on the basis of the direct method of parametrical identification, i.e. by substitution into formula (4) of a sequence of the values ρ_{gr1} and P, taken from the Table, with subsequent solution of the obtained systems of equations.

As one can see from the Figure, all experimentally obtained points are located rather close to the curve built according to formula (4). This proves that mathematical model (4) allows getting rather correct idea about character of the real dependence of the specific electric resistance ρ_{gr} upon pressure *P*. It should be noted that analytical expression (4) is, due to its simplicity, rather convenient for mathematical description and computer modeling of electric processes, proceeding in the granular environment.

CONCLUSIONS

1. It is established that specific electric resistance of the granular material is practically invariant relative size of the grains, but significantly depends upon external pressure.

2. It is shown that the idea of specific electric resistance of the granular material itself without a specifying reservation concerning the level of applied external pressure on the granular material loses any sense.

3. It is determined that in all cases, when specific numerical values of specific electric resistance of the granular material are meant, it is necessary to obligatory indicate external pressure, at which these values were obtained.

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CONGRATULATE WITH JUBILEE!

Gennady F. Torkhov ---- prominent specialist in the field of metallurgy of high-purity metals and alloys, candidate of technical sciences, winner of State Prize of Ukraine ---- is seventy.

G.F. Torkhov was born on 24th of January 1937 in village Kamenny Klyuch (Udmurtiya, RF). He started his work at Izhevsk metallurgical plant after graduation with honors from industrial technical school. He got his higher education in Izhevsk Mechanical Institute and S.M. Kirov Ural Polytechnic Institute, from which he graduated with honors in 1963. Being grant-aided student of the plant, G.F. Torkhov returned to Izhevsk metallurgical plant, where he occupied position of leader of the group for production of high-alloy steels and alloys.

G.F. Torkhov has been working in the E.O. Paton Electric Welding Institute since 1966 ---- first as a post-graduate student, then junior, senior scientific worker, and head of the laboratory.

In 1970 G.F. Torkhov defended candidate's thesis and got three-month on-job training in Ohio University, USA. In 1976 he was given scientific grade of senior scientific worker. The main scientific field, in which he worked, was development of plasma-arc processes for production of high-quality metals and alloys.

G.F. Torkhov actively attracts talented graduates of higher educational institutions for development in PWI of new metallurgical technologies. With his direct assistance such known specialists as V.K. Granovsky, Yu.I. Kostenko, V.K. Kedrin, M.I. Taranov and A.V. Likhobaba fulfilled original investigations and defended candidate's theses.

G.F. Torkhov won State Prize of UkrSSR in the field of science and technology in 1980 for development of fundamentals and design and introduction into industry of the technology and the equipment for plasma-arc melting of steel ingots and alloys from billets and non-compact charge.

G.F. Torkhov is well known and respected at the plants of CIS quality metallurgy, where high-effi-

70th birthday 50-year working record 40 years in PWI



ciency plasma metallurgical processes, among which should be especially noted alloying of steel by nitrogen from gas phase and plasma-arc remelting of surface layer of ingots and billets ---- a waste-free technology in metallurgy that is widely used in the industry, were implemented with his direct participation. In 1990 at the plants «Dneprospetsstal», Zaporozhie, Ukraine, and «Elektrostal», Electrostal, RF, the main bulk of electrodes of special VAR alloys were processed using this technology.

G.F. Torkhov participated many times in international scientific conferences, at which he worthily presented developments of the E.O. Paton Electric Welding Institute and Ukrainian science.

Scientific works of G.F. Torkhov include 197 articles, 105 copyright certificates and patents for inventions, and two books.

We cordially congratulate Gennady F. Torkhov with the anniversary and wish him every creative success and good health.

> E.O. Paton Electric Welding Institute Editorial boards of the journals «Advances in Electrometallurgy» and «Sovremennaya Elektrometallurgia»