

Editor-in-Chief B.E. Paton

Editorial Board:

D. Ablitzer (France)
D. I. Dyachenko
exec. secr. (Ukraine)
J. Foct (France)
Ö. El Gammäl (Germany)
I. J. Gasik (Ukraine)
G. I. Grigorenko
vice-chief ed. (Ukraine)
B. Eoroushich (Slovenia)
V.I. Lakomsky (Ukraine)
V.É. Lebedev (Ukraine)
S.F. I edina (Spain)
L.B. I ädivär (Ukraine)
A. I itchel (Canada)
B.Ä. I vchan (Ukraine)
Ä.N. Petrunko (Ukraine)
Ts.V. Räshäv (Bulgaria)
N.P. Örigub (Ukraine)
A.A. Troyansky (Ukraine)
I.L. Zhadkevich (Ukraine)

Executive director

A.T. Zelnichenko

Translator

S.A. Fomina

Editor

N.A. Dmitrieva

Electron galley

I.S. Batasheva,

T.Yu. Snegiryova

*Editorial and advertising offices
are located at PWI:*

International Association «Welding»,
E.O. Paton Electric

Welding Institute of the NASU,
11, Bozhenko str., 03680,
Kiev, Ukraine

Tel.: (38044) 287 67 57,
529 26 23,

Fax: (38044) 528 04 86

E-mail: journal@paton.kiev.ua

http://www.nas.gov.ua/pwj

Subscriptions:

4 issue per year;

184\$ — regular, **150\$** — for subscription

agencies, **100\$** — for students;

postage and packaging included.

Back issues available.

All rights reserved.

This publication and each of the articles
contained herein are protected by copyright.
Permission to reproduce material contained in
this journal must be obtained in writing from
the Publisher.

Copies of individual articles may be obtained
from the Publisher.

CONTENTS

ELECTROSLAG TECHNOLOGY

Kompan Ya.Yu., Nazarchuk A.T. and Protokovilov I.V. Towards the
problem of intensification of electromagnetic action in
magnetically-controlled electroslag melting of titanium alloys 2

Tsykulenko K.A. Progress of electroslag technologies and
updating of designs of ESR moulds (Review) 7

ELECTRON BEAM PROCESSES

Ustinov A.I., Fesyun E.V., Melnichenko T.V. and Romanenko S.M.
Effect of substrate temperature on micro- and substructure of
copper condensates deposited from a vapor phase 18

PLASMA-ARC TECHNOLOGY

*Shapovalov V.A., Nikitenko Yu.A., Burnashev V.R. and Torkhov
G.F.* Measurement of temperature of drum-cooler in the process
of super-rapid hardening of metal melted in plasma furnace 25

GENERAL PROBLEMS OF METALLURGY

Khizhnyak V.G., Pomarin Yu.M., Kurilo N.A. and Medova I.Yu.
Diffusion coatings on the base of titanium, vanadium and
chromium carbides on steel U8A 28

*Malashenko I.S., Kurenkova V.V., Belyavin A.F., Trokhimchenko
V.V., Chervyakova L.V. and Lebedevich Ya.B.* Structure and
strength properties of brazed joints of cast nickel alloy JS26NK.
Part 1 31

Postizhenko V.K. and Efimenko G.G. Analysis of the
state-of-the-art and problems of development of continuous
casting of steel in metallurgy of Ukraine 39

ELECTROMETALLURGY OF STEEL AND FERROALLOYS

*Panchenko A.I., Salnikov A.S., Logozinsky I.N., Korol L.N.,
Kazakov S.S., Gasik M.I. and Gorobetz A.P.* Comparative
experimental-industrial investigations of effect of 65 %
ferrosilicium with different content of calcium on contamination of
steel ShKh15SG-V with globular alumocalcium inclusions 46

REVIEWS AND SCIENTIFIC DISCUSSIONS

Lakomsky V.I. Concerning the article of G.A. Tsybulkin «On issue
of dependence of the carbon granular material electric properties
upon pressure and size of fractions» 53

INFORMATION

Repair brazing of parts of the turbine hot path 56

Developed at PWI 27, 38, 45, 55

Index of articles for AEM'2007, Nos. 1–4 57

List of authors 58



TOWARDS THE PROBLEM OF INTENSIFICATION OF ELECTROMAGNETIC ACTION IN MAGNETICALLY-CONTROLLED ELECTROSLAG MELTING OF TITANIUM ALLOYS*

Ya.Yu. KOMPAN, A.T. NAZARCHUK and I.V. PROTOKOVILOV

E.O. Paton Electric Welding Institute, NASU, Kiev, Ukraine

The effect of a longitudinal constant magnetic field on technological and metallurgical peculiarities of the process of magnetically-controlled electroslag melting of titanium alloys was investigated. It was shown that the use of a pulsed action makes it possible to refine the structure of melted ingots, preserving the high quality of their formation. Conditions of a pulsed electromagnetic action for melting titanium ingots of 80–140 mm diameter have been developed.

Key words: magnetically-controlled electroslag melting, longitudinal magnetic field, hydrodynamics, ingot, titanium alloys, macrostructure

Growing requirements to the service properties and life of titanium products demand the development of new metallurgical compositions of titanium alloys and technological processes of their production. Moreover, the chemical and structural homogeneities of metal are the most important characteristics defining the service properties of the products; their improvement can be considered as one of the basic trends in the development of the melting technology.

In this connection, the investigation of possibilities of improving the quality of metal melted by the effect of external magnetic fields on the process of crystallization of ingots metal is of a special interest [1–3]. The technology of the magnetically-controlled electroslag melting (MEM) of titanium alloys, developed at the E.O. Paton Electric Welding Institute, was based on this approach [2–3]. The application of an electromagnetic control of hydrodynamics of metallurgical melt in MEM makes it possible to influence the main processes of the metallurgical processing, namely melting of a consumable electrode, transfer of electrode drops through a slag pool, ingot crystallization. Thus, there appears a feasibility to activate the processes of chemical interaction of slag and metal, to control the direction and intensity of flows of slag and molten metal, shape and volume of a molten metal pool and, finally, to produce the chemically- and physically-homogeneous ingots with a fine-grain structure.

In the present work, the feasibility of effect of an external longitudinal magnetic field on the process of MEM of titanium alloys was studied. To improve the effectiveness of a hydrodynamic effect on the molten

metal, a pulsed action by a constant magnetic field was used.

The magnetic field is longitudinal, where the field lines in the zone of melting are oriented mainly parallel to the axis of a consumable electrode. It can be generated by a solenoid, embracing the mould. This scheme, unlike the scheme of MEM in a transverse field, allows eliminating the application of massive magnetic conduits for intensification and concentration of a magnetic field and avoiding the residual magnetizing of the circuit.

In case of using the longitudinal magnetic field in a cylindrical system of coordinates (z, r, φ) (Figure 1) the current of melting has components $I_z, I_r, 0$, and magnetic field — $B_z, 0, B_\varphi$, where B_φ is the natural magnetic field of melting current; B_z is the magnetic field of a solenoid current. Then, the electromagnetic force $\vec{F}_e = \vec{I} \times \vec{B}$ will have components $\vec{F}_e = (I_r B_\varphi - I_z B_\varphi, I_r B_z)$. It is seen from this equation that meridian (in plane $z-r$) flows are created only as a result of interaction of melting current with a natural magnetic field, while an external longitudinal magnetic field in interaction with a radial component of melting current creates an azimuthal electromagnetic force, rotating the molten metal around the axis of symmetry (Figure 1, pos. 5). Moreover, the direction of the azimuthal rotation of the molten metal depends on a sign of the external magnetic field, i.e. by changing the current direction in solenoid windings it is possible to realize the reversing movement of the molten metal.

It should be noted that the above-mentioned statement is justified both for the slag and also metal pools. However, the intensity of the azimuthal rotation of a molten metal will be lower than that of a slag, that is due to the nature of current distribution

*In this work engineers A.N. Petrov, D.A. Petrov, A.M. Belov took part.



in the metal pool and to thermal-physical characteristics of metal and slag.

Thus, the interaction of a longitudinal magnetic field with a radial component of melting current generates the metal rotation around the pool axis. But, as was mentioned in work [4], this rotation promotes the increase in a metal pool depth and undesirable change in its shape. Moreover, the growth of crystals approaches to radial one, that reduces the effectiveness of metal purification from impurities and promotes the development of liquation processes in the ingot.

It should be outlined that the rotation of melts is generated at alternating and constant values of current of melting and external magnetic field. In case of using the constants of external magnetic field and alternating current of melting the direction of action of the electromagnetic force is changed into opposite one at 50 Hz frequency of the melting current. This results in molten metal vibration (Figure 2), and the electric eddy flows in vertical planes are retarded. The mentioned effects (molten metal vibration and reduction in intensity of traditional flows, directed downward on the pool axis) can represent a special interest for the technology of melting. In addition, a positive moment of this method of an electromagnetic action is the absence of need in manufacture of special split moulds for eliminating the effect of a short-circuited loop.

The effect of a constant longitudinal field on parameters of MEM process and metal macrostructure was studied. Consumable electrodes of 50 mm diameter, made from commercial titanium VT1, were remelted into 80 mm diameter ingots. Melting was performed in a chamber-type furnace in argon using the alternating current. The level of induction of external magnetic field varied within 0.05–0.45 T ranges. Here, both a constant and pulsed action of the magnetic field on the crystallizing molten metal was used.

The observation of a free surface of the slag pool showed that the interaction of an alternating current of melting with a constant longitudinal magnetic field creates the molten metal rotation, except vibration, around a pool axis, stipulated, probably, by the presence of a constant component of current of melting owing to its partial rectifying [5].

The carried out investigations showed that the longitudinal magnetic field at induction of up to 0.1 T does not have a great influence on the melting conditions and structure of ingot being melted. With increase in induction up to 0.14–0.18 T a disorientation of a crystalline structure of the ingot was observed, while at induction of 0.2–0.4 T its refining was observed (Figure 3, *b*). However, the action of the magnetic field at induction of more than 0.2 T leads to the noticeable deterioration in the quality of a lateral surface of the ingot and appearance of slag inclusions at the ingot periphery (Figure 3). Thus, the application of the above-mentioned method did not allow optimizing the effect of MEM process on

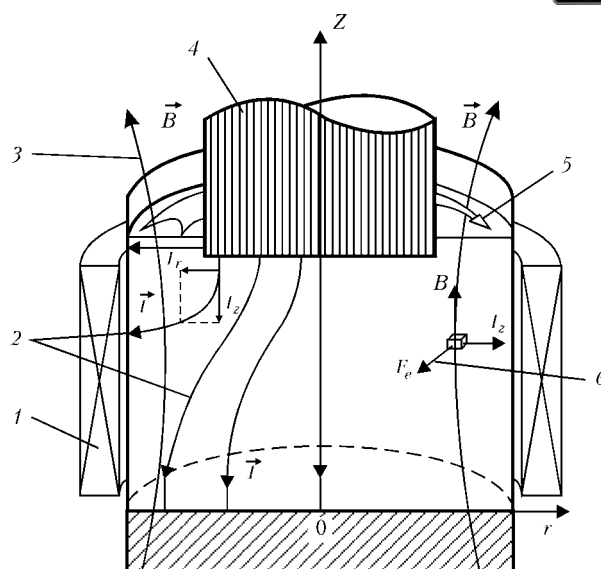


Figure 1. Mode of a slag melt rotation during melting in a longitudinal magnetic field: 1 — solenoid; 2 — lines of electric current; 3 — magnetic field force lines; 4 — electrode; 5 — azimuthal rotation of the melt; 6 — electromagnetic force

ingot crystallization and guaranteeing the acceptable quality of its surface. At a favorable action of the magnetic field on the ingot crystallization the deterioration of the quality of its lateral surface was observed as a rule, even with the formation of slag inclusions and other defects of formation.

To increase the effectiveness of a hydrodynamic action on the molten metal with preserving of the high quality of formation of the ingot lateral surface, the method was used, providing a pulsed action of the longitudinal magnetic field. The pulsed magnetic fields promote a constant restructuring of hydrodynamic flows in a metallurgical pool, providing intensive stirring of the molten metal. At the same time they do not allow forming of stable eddy structures



Figure 2. Vibration of a model liquid (concentrated solution of CaCl_2) under the action of longitudinal constant magnetic field at $I = 40 \text{ A}$, $B = 0.2 \text{ T}$

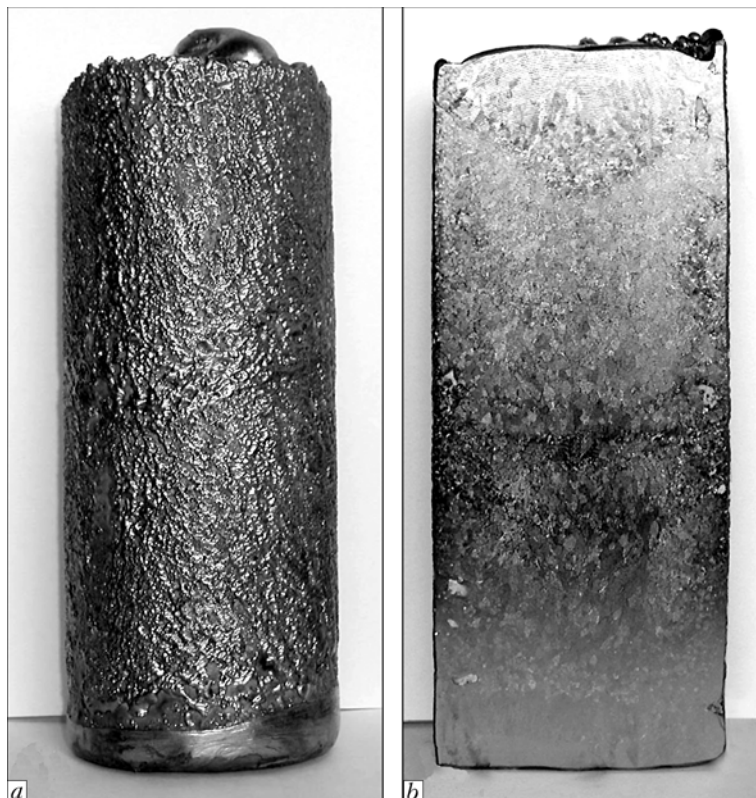


Figure 3. Appearance (a) and macrostructure (b) of ingot melted under the action of a longitudinal magnetic field of 0.24 T induction

in a metal pool, which can lead to different kinds of chemical heterogeneity of the ingot.

To generate a pulsed longitudinal field, the solenoid was supplied by separate DC pulses. Duration of pulses t_{pulse} was preset within 0.3–33 s, t_{pause} — 3–33 s.

It was established that at a stationary action of the magnetic field the melting current is decreased at the moment of this field switching on, and after a short period of time it is recovered up to the value which is somewhat lower (by 10–20 %) than the initial value. Experiments with a pulsed action of the longitudinal magnetic field showed a regular phenomenon, where the pulsed action of the longitudinal magnetic field causes the appropriate pulsed change in melting current without a purposeful change in melting voltage and speed of the consumable electrode feed (Figures 4 and 5).

At the moment of magnetic field switching on, the abrupt drop in melting current is occurred down to a definite minimum of value defined, first of all, by the level of induction of the external longitudinal magnetic field, duration of its action and melting

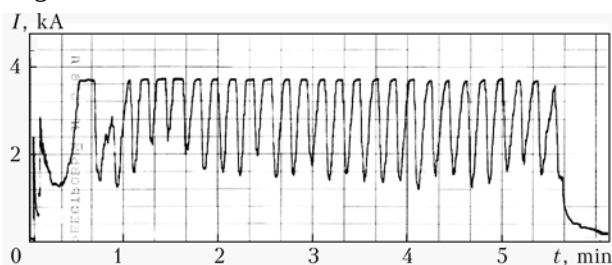


Figure 4. Registogram of melting current at a pulsed action of a longitudinal magnetic field ($t_{\text{pulse}} = 1$ s, $t_{\text{pause}} = 10$ s, $B = 0.21$ T)

current (Figure 5). At induction 0.21 T and pulse duration of more than 0.8 s, the melting current was reduced by 40–60 % (Figure 4). At a short duration of the magnetic field pulse (less than 1.5–2.0 s) the melting current after its abrupt drop is recovered quite quickly by the self-control of the electroslag process up to the initial values (Figures 4 and 5, a). In case of a longer pulse of the electromagnetic action the melting current began to grow already during the action of the magnetic field (Figure 5, b) and at a sufficient duration of the pulse (of more than 15–20 s) it is stabilized at a definite level, somewhat lower (by 10–20 %) than the melting current without the electromagnetic action (Figure 5, c). Then, when the magnetic field was switched off, the current increased up to the initial value.

In principle, the mechanisms of melting current decrease under the action of the longitudinal field are similar to the mechanisms of a current change under the action of a transverse field [3]. These mechanisms depend on decrease in a depth of electrode immersion into a slag pool, increase in an interelectrode gap and increase in electric resistance of consumable electrode–metal pool region of the circuit.

Thus, the pulsed magnetic field causes the discrete-portion heat input, creating additional capabilities for the control of the ingot crystallization process and metal quality.

Metallographic examinations of ingots showed that the pulsed magnetic fields are an effective means for control of the structure of metal being melted. Mechanism of ingot structure refining is connected with a fracturing action to growing crystals using the

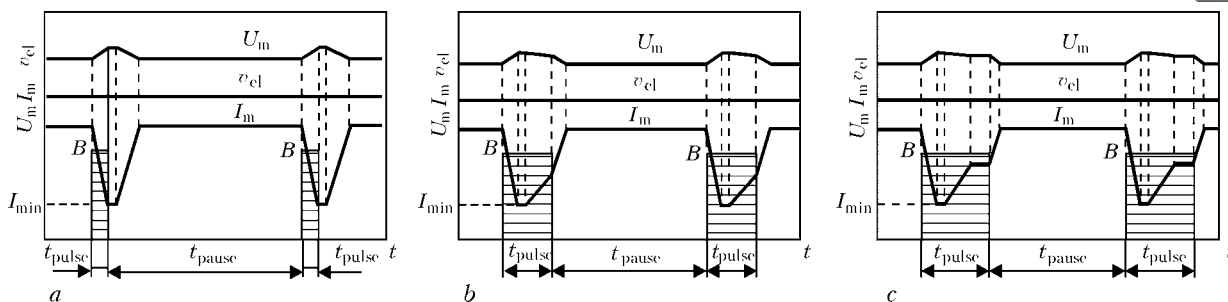


Figure 5. Schemes of changing the electric current in MEM at a pulsed action of a longitudinal magnetic field in cases, when the duration of a pulse is not sufficient (*a*, *b*) and sufficient (*c*) for stabilization of melting current during a separate pulse: *B* — induction of external magnetic field; U_m — melting voltage; v_{el} — electrode feeding speed; I_m — melting current; t_{pulse} , t_{pause} — duration, respectively, of a pulse and pause of action of the external magnetic field

electromagnetic vibration of the molten metal, caused by the interaction of melting alternating current with external constant magnetic field. This mechanism is amplified by magnetohydrodynamic impacts at the moment of a cyclic change in induction of the longitudinal magnetic field. The ingot structure refining is also due to the intensification of a heat-mass exchange at the liquid–solid phase boundary.

Comparison of obtained data about the effect of pulsed magnetic fields on the electroslag process conditions and structure of ingots being melted showed that the energy of the electromagnetic action at a preset melting current is determined by a duration of action and level of magnetic induction. To provide a sufficient induction of the magnetic field, there are upper and lower limits of duration of action pulses and pauses, influencing the quality of surface, disorientation and refining of the crystalline structure.

At a pulse duration of less than 0.8 s and induction of magnetic field $B = 0.2\text{--}0.3$ T, the duration of a hydrodynamic action on the molten metal is not sufficient for the effective refining of the macrostructure of ingots. With increase in duration of pulses of mag-

netic field action the tendency to the more noticeable refining of the ingot crystalline structure is observed. At the same time, with increase in duration of the pulsed action the quality of formation of a lateral surface of ingots is decreased. In melting with a long-time action of the field (more than 5 s) a non-uniform formation of ingot surface with interceptions, occurring as a result of action of the external magnetic field, was observed (Figure 6).

The quality of the ingot lateral surface formation is also deteriorated greatly at decrease in duration of pauses of the magnetic field (less than 4 s). With increase in duration of pauses the quality of the surface of ingots is improved, but the effectiveness of metal macrostructure refining is reduced, especially at insufficient induction of the external magnetic field or a low duration of its action.

Thus, it is possible to recommend the following conditions of a pulsed action by a longitudinal magnetic field for melting ingots of 80 mm diameter: $B = 0.2\text{--}0.3$ T; $t_{pulse} = 1\text{--}2$ s; $t_{pause} = 8\text{--}15$ s. Here, with increase in magnetic field induction the duration of its action can be somewhat reduced. Ingots, melted by using these conditions of electromagnetic action,



Figure 6. Appearance of ingot melted under the action of a pulsed longitudinal magnetic field ($t_{pulse} = 11$ s, $t_{pause} = 21$ s, $B = 0.21$ T)

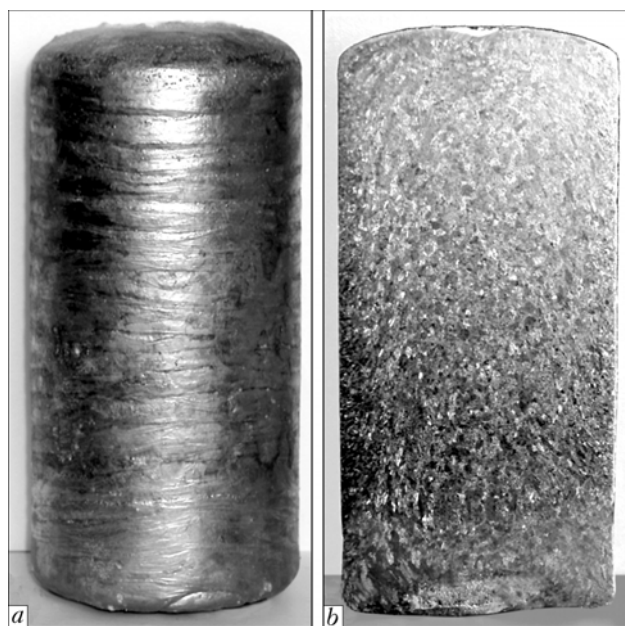


Figure 7. Appearance (*a*) and macrostructure (*b*) of ingot melted under the action of a pulsed longitudinal magnetic field ($t_{pulse} = 1$ s, $t_{pause} = 10$ s, $B = 0.21$ T)



Mechanical properties of high-alloy titanium alloys melted at a pulsed action of the longitudinal magnetic field

Alloy	$\dot{\sigma}_{\text{test}}, \text{ }^\circ\text{N}$	$\sigma_{0.2}, \text{ MPa}$	$\sigma_t, \text{ MPa}$	$\delta, \%$	$\psi, \%$
VT22 + 5 % Sn + 0.45 % Si	20	1310	1360	3	8
	600	380	460	52	90
VT22 + 5 % Cr + 0.3 % Si	20	1220	1235	11	17
	600	350	430	60	90
Ti – 4.5 % Al – 25 % Nb – 5.0 % Mo – – 4.0 % Fe – 0.1 % B – 0.1 % C – 0.1 % Si	20	1318	1380	3,5	5,5
	650	480	525	36	72

are characterized by a homogeneous fine-grain structure (1.0–1.8 mm grain size) without slag inclusions and other defects of formation and a good quality of a lateral surface (Figure 7). The rather high stability of refining of crystalline structure of ingots should be especially outlined in this case.

As to the assessment of effectiveness of action of pulsed longitudinal magnetic fields on the process of melting of large-diameter ingots of high-alloy titanium alloys, then the additional investigations are required.

At a final stage of investigations, the experimental melting of 140 mm diameter ingots of alloy VT1 was made. Induction of magnetic field in the mould center was 0.34–0.40 T, $t_{\text{pulse}} = 1.5\text{--}3.0$ s, $t_{\text{pause}} = 7\text{--}14$ s. Results of experiments confirmed the effectiveness and reliability of the developing method of intensification of electromagnetic action in magnetically-controlled electroslag melting.

Melting of a number of experimental titanium alloys with an intermetallic type of strengthening was also performed. Compositions of investigated alloys were developed on the basis of ($\alpha + \beta$)-alloys, additionally alloyed by eutectoid-forming elements for improving the heat resistance of metal, for example, by formation of chemical compounds in a solid-solution matrix [6–7].

To produce ingots of appropriate compositions, the 45 mm diameter consumable electrodes, extruded from spongy titanium and alloying components, were used. Two-time remelting was used. At the second stage the melting was performed by a pulsed action of the longitudinal magnetic field using the worked out recommendations. The produced metal was subjected to proper thermal mechanical treatment and heat treatment. Results of tests of this metal are given in the Table. They prove the feasibility of producing high-alloy titanium alloys with intermetallic type of strengthening, characterized by high values of strength characteristics, preserving the required level

of ductility, using the method of MEM in pulsed magnetic fields.

CONCLUSIONS

1. It is shown that the external pulsed magnetic field is an effective means of control of properties of metal melted in MEM of titanium alloys. Its use allows refining macrostructure of ingots, preserving the high quality of their formation.

2. It was established that the action of a pulsed longitudinal magnetic field causes a cyclic change in melting current and, respectively, discrete-portion heat input in a slag pool, thus creating the additional capabilities for control of the ingot crystallization process.

3. The conditions of a pulsed action by a longitudinal magnetic field have been developed for melting titanium ingots of 80–140 mm diameter. The developed method has been tested successfully in melting of a number of experimental high-alloy titanium alloys with intermetallic type of strengthening.

1. Chernyavsky, V.B., Ladokhin, S.V., Gladkov, A.S. (2005) Melting crucibles and moulds with systems of electromagnetic stirring for electron beam melting of metals and alloys. *Protsessy Litiya*, **1**, 48–55.
2. Kompan, Ya.Yu., Protokovilov, I.V. (2002) Some technological aspects of magnetically-controlled electroslag melting (MEM) of titanium alloys. In: *Special Metallurgy: Yesterday, Today, Tomorrow: Proc. of Int. Sci.-Techn. Conf.* (8–9 Oct. 2002, Kiev), 256–262.
3. Protokovilov, I.V. (2006) *Magnetically-controlled electroslag melting of multicomponent titanium alloys*: Synopsis of Thesis for Cand. of Techn. Sci. Degree. Kyiv.
4. Paton, B.E., Medovar, B.I., Us, V.I. et al. (1989) Development and study of methods of control of solidified ESR ingot structure by superposition of magnetic field. *Problemy Spets. Elektrometallurgii*, **4**, 3–7.
5. Maksimovich, B.I. (1961) About alternative current rectification in electroslag remelting of consumable electrodes in water-cooled metal mould. *Avtomatich. Svarka*, **3**, 101–102.
6. Kompan, Ya.Yu., Protokovilov, I.V., Moiseev, V.N. (2004) High-strength and heat-resistant titanium alloys with intermetallics produced by MEM technology. *Advances in Electrometallurgy*, **2**, 2–6.
7. Sysoeva, N.V., Moiseev, V.N. (2002) Titanium alloys with intermetallic type of strengthening. In: *Aircraft materials and technologies*. Moscow: VIAM.



PROGRESS OF ELECTROSLAG TECHNOLOGIES AND UPDATING OF DESIGNS OF ESR MOULDS (Review)

K.A. TSYKULENKO

E.O. Paton Electric Welding Institute, NASU, Kiev, Ukraine

Different types of designs of moulds, as well as the methods of producing ingots and castings, both used only in lab practice and also widely in industry, are considered. The updating of designs is described depending on various demands of the electroslag production. Advantages of electroslag processes in a current-carrying mould, problems occurring at the present stage, and also possible trends of development are shown.

Key words: electroslag technologies, designs of moulds, ingots and castings, separation of processes of melting and crystallization, possible trends in the development of electroslag technologies

With a progress of the electroslag process as a method of producing high-quality ingots and castings the designs of moulds were updated. At the first stages, when electroslag remelting (ESR) was only began to be developed on the basis of the electroslag welding, and melting of ingots was realized using an electrode wire, the simple water-cooled vessels of a round section, as a rule, were used as moulds for the producing ingots [1]. In use of the electrode wire during ESR for producing metal of a preset chemical composition a bulky master alloy was supplied into a slag pool. However, this method of ESR did not find industrial application because of a high cost of the wire itself and master alloy, and also difficulty in providing the chemical homogeneity of the ingot. The use of ready electrodes of a preset chemical composition for remelting, as well as increase in the ingot section, allowed eliminating the mentioned problems, and very soon the industrial remelting began to be used using first the electrodes of small (30–50 mm) and then of large (1200 mm) diameters.

In ESR, the ingots of different cross-section are produced. Using the electrodes of a different shape of the cross-section and varying their arrangement with respect to each other, it is possible to change the shape and ratio of walls of ingots being remelted within the wide ranges. To produce these ingots, the appropriate designs of moulds have been developed, in which both forge and also slab ingots of required types and sizes can be produced (Figure 1).

In remelting into moulds of a rectangular section, corrugations can form in the area of narrow sides and angles on the ingot surfaces. This is caused by the fact that the heat in a slag pool is generated mainly in a near-electrode zone, and it is transferred into the periphery regions only owing to the heat conductivity and convection in a molten slag. To form the quality

surface, chamfers or roundings of appropriate radii can be specified on ribs of slab ingots of small and medium sizes (Figures 1, c, d). The quality of surface of the slab ingot is greatly improved in use of short moulds with a ribbed surface [2]. The presence of ribs on the working surface of a plane mould promotes the formation of a defect-free surface of the ingot being remelted, and also allows increase in area of mould and ingot contact, thus intensifying its cooling. To melt large slab ingots, the moulds with concave lateral sides of an appropriate radius are used (Figure 1, e, f). The lateral sides of these ingots are produced convex. This allows providing the necessary ratio of deformation of a large slab ingot in its next rolling, as well as minimum losses of metal for cutting [3].

In a canonical diagram of ESR the efficiency of the process and metal quality are in an inverse dependence. Therefore, it was necessary to provide the required quality in practice by decrease in ESR process efficiency, especially in producing the ingots of a small mass. A multi-passage diagram of ESR (Figure 2) [4] became a new step in the way of increasing the efficiency and also decrease in energy consumption and metal cost reduction, allowing producing several small-section ingots simultaneously in a large ESR furnace.

The further progress of the electroslag technology contributed to the development of the electroslag production of pipes. Special furnaces were designed and manufactured for this purpose. The main element in them is special moulds, design of which is defined to a larger extent by a variant of the remelting technology. Known are the schemes of melting hollow ingots with an arrangement of electrodes in a circular gap between the external and inner moulds. This can be scheme with use of a separate internal mould, i.e. mandrel (Figure 3, a) or scheme, in which external and internal moulds represent an integral structure (Figure 3, b) [5]. The internal mould (mandrel) is made usually canonic. It is moved during melting as regards to ingot to prevent a jamming in metal shrink-

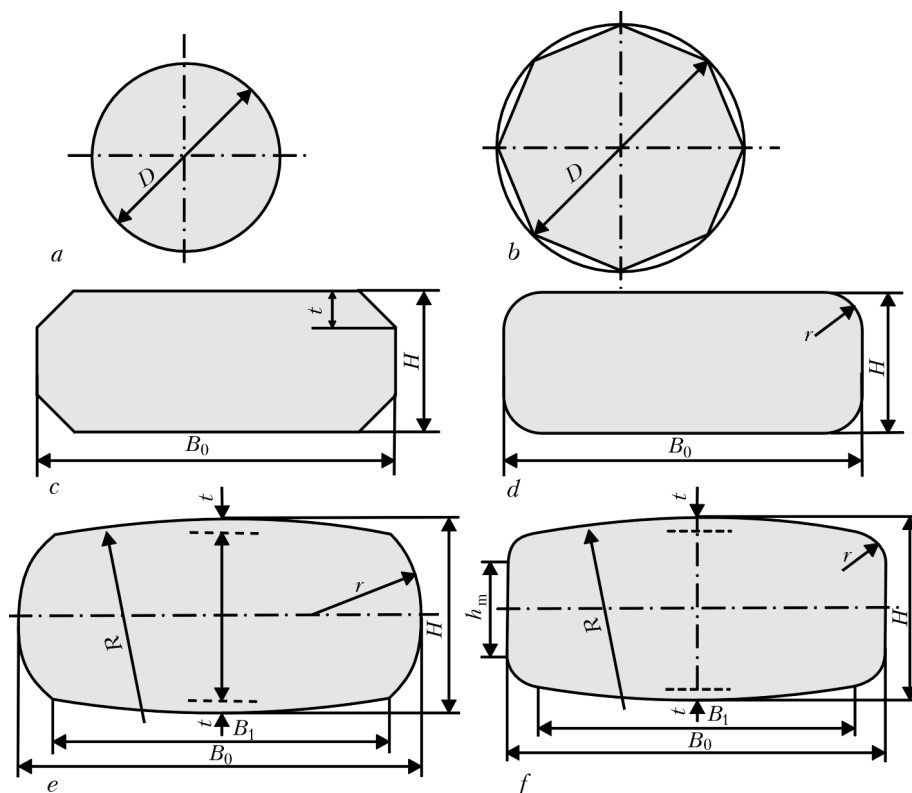


Figure 1. Sections of forming parts of some moulds for melting of forge (a, b) and slab (c-f) ingots

age. The main drawback of these schemes of remelting consists in a need of using several consumable electrodes of a small section simultaneously and, consequently, of a rather large length.

There is also a scheme of electroslag piercing of a solid electrode (Figure 3, c). This method makes it possible to use one solid electrode of a large cross-section. The main drawback of this scheme of remelting is the probability of deviation of a piercing mandrel from the ingot axis and formation of a hollow ingot of a different wall thickness, which increases with a growth in a hollow ingot length.

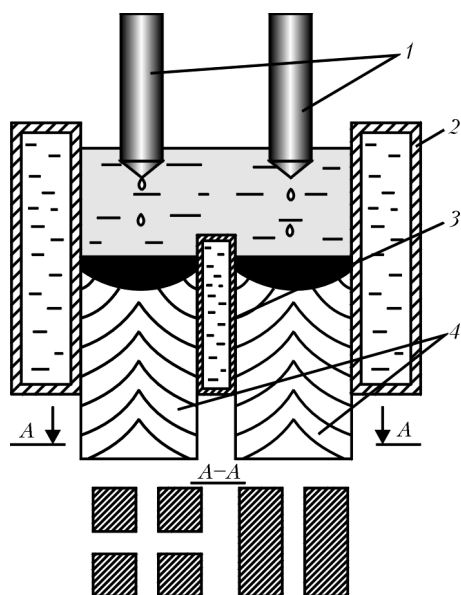


Figure 2. Scheme of a multi-pass ESR: 1 — consumable electrodes; 2 — mould; 3 — water-cooled partition; 4 — ingots

Updating of the electroslag process of production of hollow ingots allowed producing canonic ingots (Figure 4, a). Melting of these ingots is performed in a mould, the shape of which corresponds to the ingot shape, with a simultaneous piercing of a hole in it [6]. In addition, the technology of producing high-quality hollow billets with square holes (Figure 4, b) [7], as well as shaped hollow ingots (Figure 4, c) [8], has been developed. During mastering the production of hollow ingots by the ESR method, technology of manufacture of circular and tubular billets was simplified and a metal utilization factor and quality of these products were increased.

Cast electroslag metal is superior greatly to that, produced by conventional methods of casting, and in many cases it is not inferior to the forged metal by service characteristics. Therefore, the electroslag process began to be widely used for producing products of a more complicated shape (castings). This technology was called electroslag casting (ESC). The main difference of ESC and ESR consists in use of a mould in ESC, the internal cavity of which has a variable section and corresponds to external shape of the casting. Depending on the degree of complexity of the casting, several schemes of ESC are used. The consumable electrode can be melted either in the mould or in an auxiliary vessel, which is moved over the mould during the casting formation (Figure 5). ESC castings can have various shapes, close to those of a ready product, including those with one or several cavities (Figure 4, d-f) [9]. The further progress of the ESC technology allowed producing even ingots of a curvilinear shape (Figure 6, a), and also such

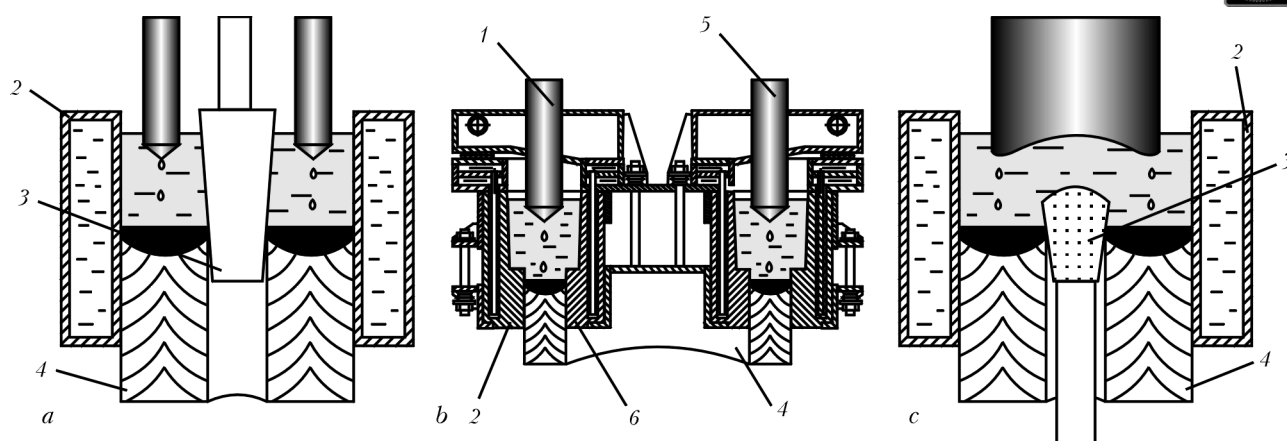


Figure 3. Scheme of melting electroslag ingots with arrangement of electrodes in a circular gap (a, b) and electroslag piercing of solid-section electrode (c): 1 — thin electrodes; 2 — external mould; 3 — mandrel; 4 — hollow ingot; 5 — solid electrode; 6 — internal mould

large shaped products, as bands of kiln furnaces (Figure 6, b, c) [10].

Design of the mould defines many parameters of the electroslag process and serviceability of the installation as a whole. Service life of the mould depends on the materials, from which it is manufactured, and conditions of its operation. It is possible to distinguish three types of mould designs: all-copper types, made from thick copper plates with bored channels in them for cooling; sleeve types (irrigated) with a copper chill mould and steel housing or without it; composite types, consisting of several panels, each of them represents a copper plate connected to a load-carrying steel basement. All-steel moulds did not find an industrial application due to often burn-through and failures, in spite of a noticeable lower cost and high

adaptability to manufacture [11]. The highest service life have all-copper stationary moulds. Movable short moulds, used in technological schemes of melting with a relative movement of the mould and ingot, are serviced in the more severe thermal conditions and their service life is lower than that of the stationary moulds operating by the scheme of filling. Melting of large-tonnage slab ingots in a short movable mould envisages a periodic replacement of damaged working panels, forming the ingots. For this purpose, different designs of the panels, including the bimetal ones, as well as methods of their sealing, have been developed [12, 13].

In stationary sleeve-type moulds of round and square sections the distortion of the structure is eliminated almost because of a uniform distribution of ther-

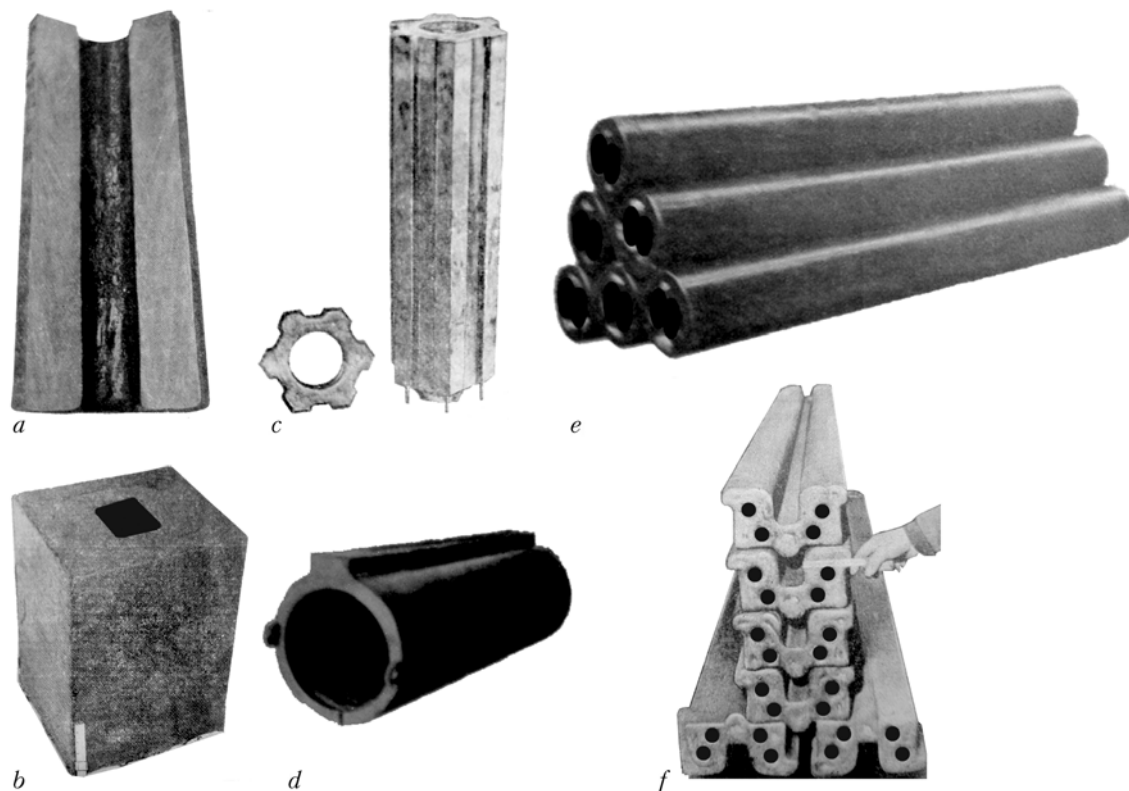


Figure 4. Electroslag hollow ingots: 1 — hollow canonic; 2 — with a square hole; 3 — shaped; 4 — with a longitudinal thickening on external surface; 5 — with internal shaped hole; 6 — long (3.7 m) shaped with four small (23 mm dia.) holes

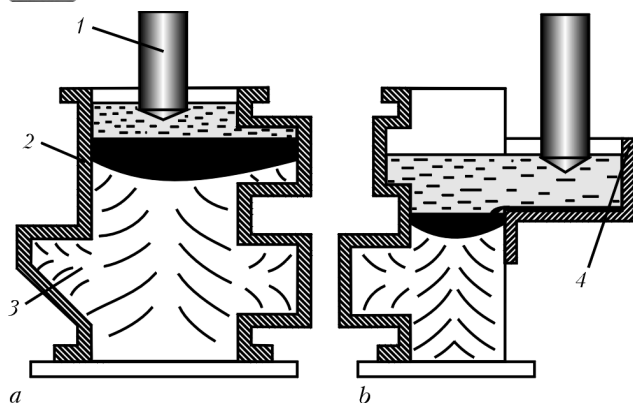


Figure 5. Scheme of electroslag casting in the mould (a) and with metal pouring-over (b): 1 — consumable electrode; 2 — mould; 3 — casting; 4 — movable melting vessel

mal load in their section unlike the moulds of a rectangular section, where the thermal load is not uniform, that stipulates the appearance of high residual deformations, mainly in the mould wide walls, which are increased from melting to melting. To produce ingots of such sections, a composite panel design of the mould is considered to be suitable.

The non-homogeneity of the temperature field influences the condition of not only the mould, but also the ingot after melting. Temperature field in hollow and shaped ingots of ESR is greatly non-homogeneous due to high rates of cooling that leads to high stresses inside the ingot. In ESR with use of a movable short mould, and also in melting of thin-walled ingots the effect of thermodeformational processes on their quality is increased noticeably. The probability of appearance of hot and cold cracks is increased, the role of shrinkage phenomena due to a possible jamming of

an internal mould in melting of hollow ingots becomes more evident.

Earlier, different measures were suggested to have a possibility to influence the conditions of cooling ingot and mould and to reduce the level of residual stresses. For example, to increase the heat transfer and increase the resistance of moulds, the working sleeve of the mould was equipped with many chambers in the form of thermal tubes (Figure 7, a) [14]; to control the process of ingot crystallization, special bands with heat-insulating elements (Figure 7, b) [15] and others were used between a water-cooled housing and a mould sleeve. However, these proposals did not find an industrial realization.

There were also attempts to change the conditions of the ingot cooling by adjustment of the cooling water temperature. As was noted in work [16], the varying in water consumption does not almost influence the conditions of the ingot cooling, because with its increase the temperature of water at the mould outlet is decreased, and the total amount of the removed heat remains the same. Methods were suggested for the ingot cooling in the mould by change of a heat carrier (instead of water the melts of salts, easily-melting metals, as well as different organic and silicon-organic liquids were used).

It was stated in work [16] that the noticeable change in conditions of the ingot cooling in the region of a movable short mould has not great influence on the residual stresses. It is possible to decrease the level of residual stresses by decrease in heat dissipation from the ingot, located already beyond the short mould, as result of warming of the ingot using different heat-insulating materials, split and telescopic

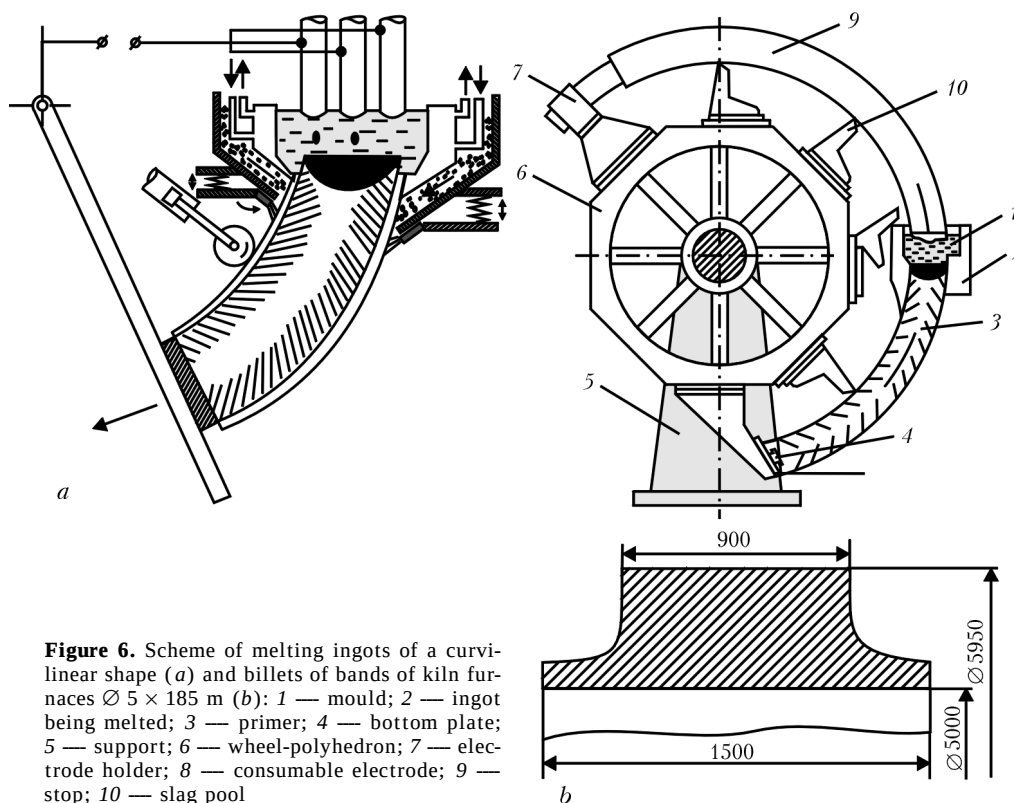


Figure 6. Scheme of melting ingots of a curvilinear shape (a) and billets of bands of kiln furnaces $\varnothing 5 \times 185$ m (b): 1 — mould; 2 — ingot being melted; 3 — primer; 4 — bottom plate; 5 — support; 6 — wheel-polyhedron; 7 — electrode holder; 8 — consumable electrode; 9 — stop; 10 — slag pool

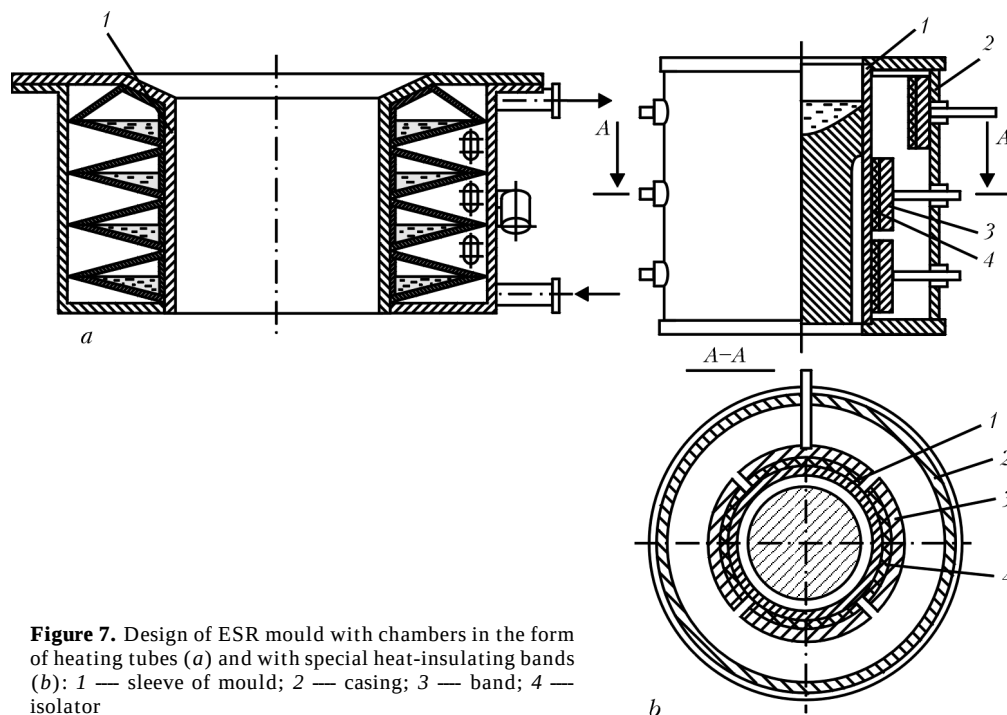


Figure 7. Design of ESR mould with chambers in the form of heating tubes (a) and with special heat-insulating bands (b): 1 — sleeve of mould; 2 — casing; 3 — band; 4 — isolator

screens, melting of ingot in a thermostat with a controllable heat removal, etc.

Moulds with differential (different in height) cooling can be used not only for adjustment of conditions of ingot cooling. A large attention is paid to a rational use of a heat generated in a slag pool.

In a classic process of ESR into water-cooled moulds about 50 % of heat, generated in a slag pool, is dissipated into the mould wall in the zone of a slag pool arrangement and only 30 % is consumed for melting electrodes and then removed by a cooling water from the ingot. It is possible to reduce the heat losses from the slag pool into the mould wall by two methods: by decrease in a slag pool depth or by increase in temperature of a mould working wall.

The depth of a slag pool is an important technological parameter. It seems impossible to decrease it below some optimum value. It is possible to increase the temperature of the mould wall by its less intensive cooling or lining with different insulating materials resistant to the action of the molten slag.

As was mentioned in work [17], the increase in temperature of the mould working wall in the slag pool zone up to 700–800 °C using the less intensive its cooling allowed decreasing the consumption of electric power for remelting and, thus, increasing the thermal efficiency factor of the installation. The temperature increase was attained as a result of the mould cooling by a boiling layer of inert particles (corundum or sand), fluidized by air. Some designs of the moulds with a lined upper part are shown in Figure 8. When melting ingots in these moulds, the slag pool is located in the upper lined section of the mould which represents an extension made of material of a low heat conductivity and temperature of melting which is higher than the temperature of the used molten slag

[18, 19]. The lined upper part of the mould can be both cooled and non-cooled.

As far back as 1959, an attempt was made to use a skull of slag, used in remelting (like skull melting) as a lining, however, it was necessary here to reduce the electric conditions of melting, that led to the deterioration of the metal quality and, sometimes, to the process interruption. When more refractory slag was used than that used for remelting, it was possible to reduce the heat losses into the mould wall from 51 (without extension) to 36 %. However the resistance of this extension was insufficient [20]. Therefore, in future the graphite or composite coatings began to be used as steel lining.

The need in control of crystallization and improvement of quality of the ESR ingots, refining of crystalline structure and prevention of formation of such defects as carbonitride liquation, layer-by-layer crystallization, light contour, contributed to the development of different methods of effect on metal during remelting (using ultrasound, electromagnetic fields, vibration, i.e. reversible displacement of mould or relative displacement of electrode and mould, etc.) and design of appropriate devices.

For example, the reversible rotation of the mould allows providing the finer grain in the ingot, more dense structure in an axial zone and decreasing the zone of columnar crystals [21]. The superposition of longitudinal reversible magnetic fields has a similar effect on crystallization of the electroslag ingot [22]. For this purpose, separate devices-inductors and inductors, arranged in a water-cooled gap between the internal sleeve and housing of the mould, can be used (Figure 9) [23]. In melting of horizontal ingots, a reciprocal movement of the mould in the horizontal plane or swinging of the mould, i.e. its symmetric inclinations for a preset angle, are used. The quality

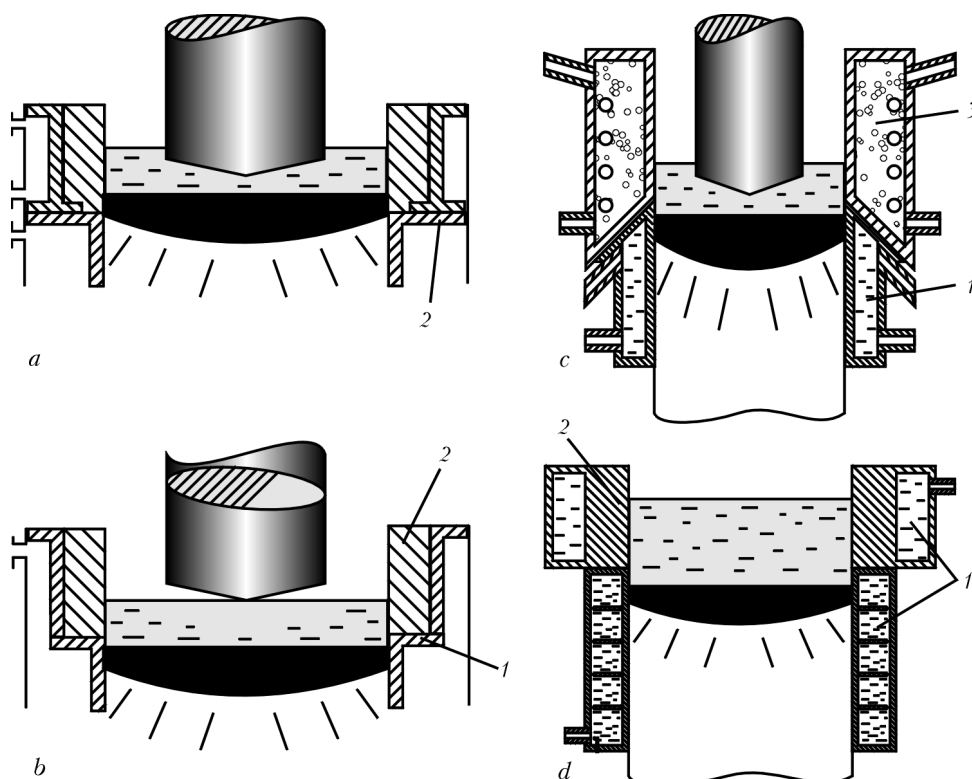


Figure 8. Designs of ESR moulds with a lined upper part (a) and differential cooling (b–d): 1 — water; 2 — heat-insulating insert; 3 — cooler with a lower heat conductivity than that of water

of metal produced by melting in the swinging mould is higher than that in melting using the scheme with a reciprocal movement [24].

In melting of heterogeneous ingots, especially in horizontal moulds, the need arises in adding of alloying materials in remelting of consumable electrodes. These materials in the form of lumps, metallic granules, chips and others are added to a slag pool using special batchers. Volume of alloying fillers can reach 100–150 % of electrode metal volume and, sometimes,

even more. Except the alloying effect, the lumpy materials can serve as an effective modifier [25]. Adding of these materials as macrocoolers, when granules are not dissolved completely and serve as additional centers of crystallization, contributes to refining of the ingot structure and increase in toughness properties of the metal.

The non-compact metallic materials serve not only as alloying or modifying additions, but also as a raw material for the melting of ingots. In melting ingots of non-compact materials the manufacture of consumable electrodes is not required. Pressing of electrodes of such charge is labor-consuming, sometimes not rational from economic point of view and not almost possible in some cases.

There are many schemes of electroslag remelting of non-compact materials. Some of them are described in work [26]. For this purpose here a new design of the mould was suggested, where the electroslag process is performed by electrodes, representing pools of molten metal and arranged on the bottom of a slag pool (Figure 10). Electrodes are isolated from each other by the cooled metal parts, covered with a crust of a solidified slag. The non-compact charge is supplied to the surface of a slag melt, moreover, the large lumps for their guaranteed melting have to be arranged on the surface of a circular electrode (on a flange of a widened mould). Earlier, the attempts were made to perform also metallothermic processes of reduction of metals in this mould [27]. Metallothermic charge together with required slag components was supplied to a mirror of a slag pool into a mould.

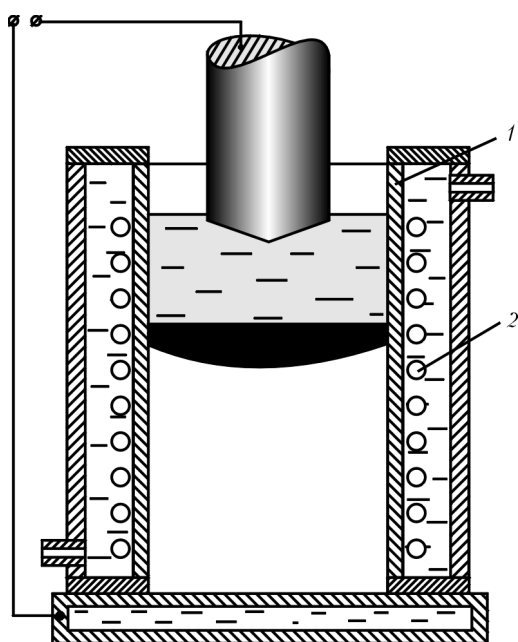


Figure 9 [23]. Method of ESR of consumable electrodes: 1 — sleeve of mould; 2 — inductor



Processes of metallothermic reduction are proceeding simultaneously with electroslag melting of metal produced as a result of reduction.

Due to insufficient resistance of elements of the mould structure, and also other problems, connected with a stable proceeding of the process, the above-mentioned scheme did not find industrial spreading, though, it was really the first scheme in which the idea of a current-carrying mould was used.

The possibility of extraction of metallic components from a slime using ESR was described in work [28]. Here, to conduct this process a copper water-cooled electrode, whose resistance attracted attention of many researchers, was used. The investigations showed that ESR can be used successfully for extraction of deficit and expensive metals from wastes. In this case, a continuous process and high purity of a metallic product from harmful impurities can be provided.

Development of new designs of current-carrying moulds and schemes of remelting in them was carried out for different purposes. Thus, need in a reliable nonconsumable electrode of a high life and at the same time maximum easy feeding of remelting (or clad) materials into a slag pool promoted the design of a sectional current-carrying mould, shown in Figure 11 [29]. The suggested design includes a device providing a slag pool rotation. Rotation is envisaged, first of all, for the temperature averaging in the slag pool volume, and also for prevention of microarcs at the slag–nonconsumable electrode boundary, that prevents a quick wear of the latter. Moreover, the surface of an upper current-carrying section, contacting a slag, is protected by a changeable heat-resistant lining which can be manufactured of graphite or refractory metal, for example, tungsten or molybdenum.

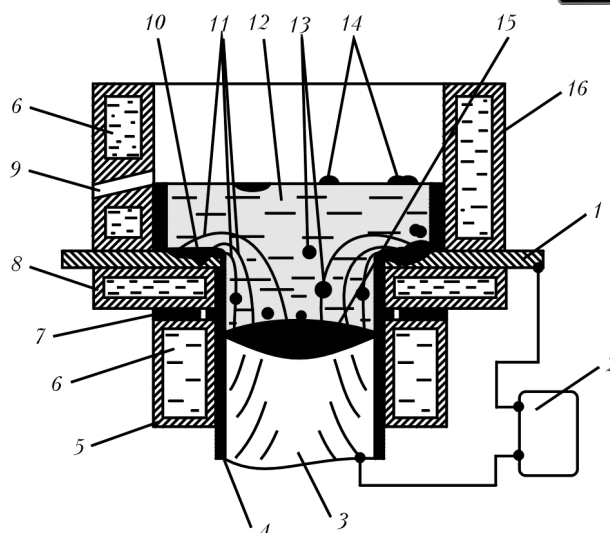


Figure 10 [26]. Electroslag furnace: 1 — circular electrode; 2 — power source; 3 — ingot; 4 — slag skull; 5 — mould; 6 — cooling water; 7 — isolator; 8 — separating element; 9 — slag tap; 10 — circular pool of molten metal; 11 — current passing paths; 12 — slag pool; 13 — molten metal drops; 14 — charge; 15 — molten metal pool; 16 — cooled wall

One of schemes of the electroslag melting of different types of stop valves using a current-carrying mould is given in Figure 12 [30]. The working surface of the mould has slots in the upper part, into which current-carrying plates are inserted (Figure 12, a), and the working surface of a mandrel has current-carrying plates for cavity formation in a casting (Figure 12, b).

The need in distribution of heat sources, leveling of a temperature field in a slag pool, as well as elimination of mould dividing into separate water-cooled sections became the basis in the development of mould designs with a distributed heat generation in a slag pool (Figure 13) [31]. Nonconsumable electrodes, iso-

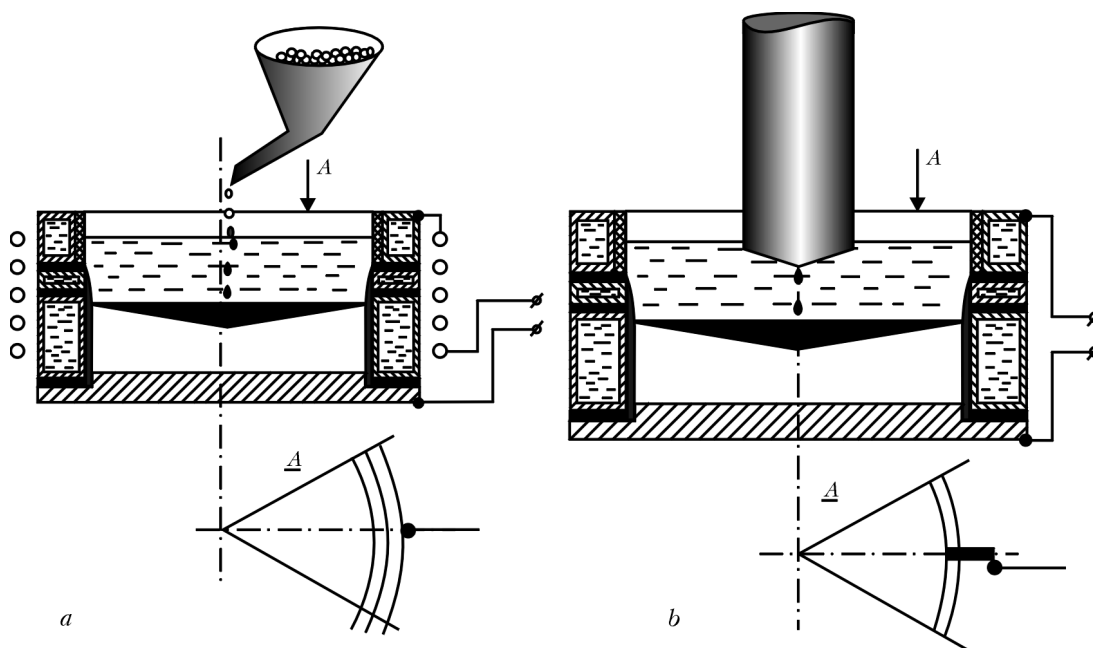


Figure 11 [28]. Scheme of current-carrying mould providing rotation of a slag pool: a — with an external inductor; b — with a split upper current-carrying sector (ring)

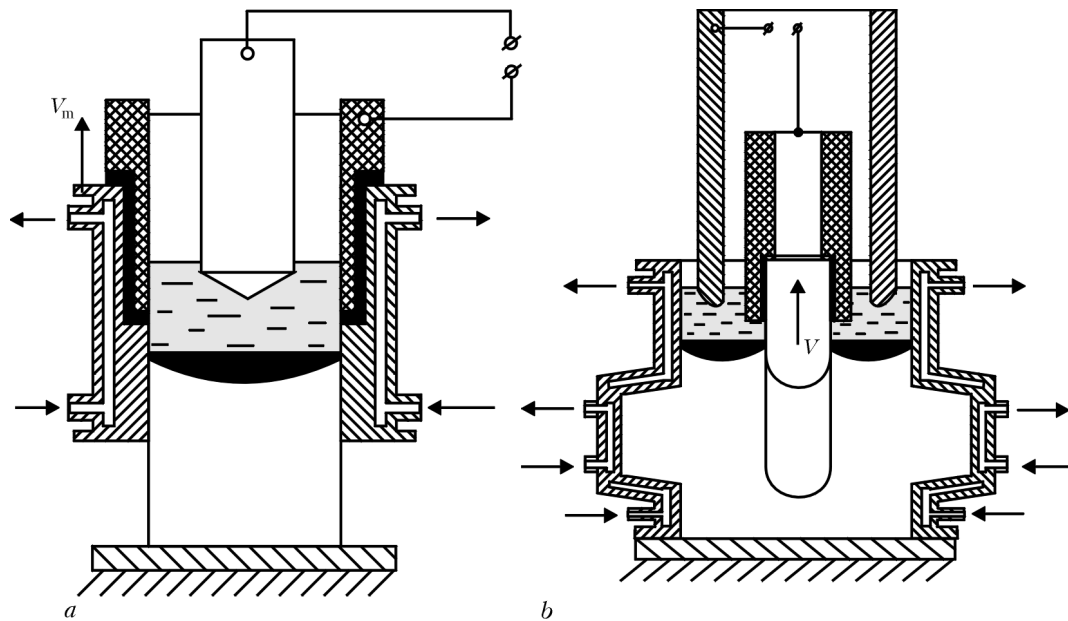


Figure 12 [30]. Installation for electroslag melting: *a* — with current-carrying plates on the mould; *b* — with a current-carrying mandrel

lated from the mould, can be arranged on its internal wall (Figure 13, *a*) or passed directly through it (Figure 13, *b*). In both cases the non-consumable electrodes are connected to the power source by a bifilar scheme. It should be noted that this scheme can operate only at a small interelectrode gap, otherwise, the slag resistance will not allow current passing even at high values of open-circuit voltage of the power source. Authors of work [31] state that as «the metal being remelted is supplied into a slag pool, overheated by these electrodes, having no electrical connection to power source and electrodes, then the rigid relation between the value of current and melting speed is eliminated completely». This statement is quite correct if a molten metal is used as a filler metal. In case of using a solid filler metal the speed of its melting and size of fraction depend on power, generated in a slag pool. As to the process, presented in Figures 11 and 13, then it is possible only to say here about the weakening of relation between the remelting current and speed of the ingot formation, rate of its crystal-

lization at given cooling conditions, as the intermediate link, i.e. metal melting speed, remains to be related with remelting current in these moulds.

The attempts were made long ago to break the rigid relation between processes of melting of filler (electrode) metal and formation of ingot (in remelting) or weld (in welding). As far back as 1941 academician V.P. Nikitin proposed to break the similar relation in fusion welding and to separate the thermal processes of preparation of parent and filler metals. Using this idea, a new method of metal joining with different methods of melting was created [32, 33]. The principle of this method consists in that the concentrated heat source heats and melts the parent metal surface for a preset depth, while a molten filler, prepared in a separate melting unit, is supplied into a pool of the molten parent metal. Thus, a noticeable reduction in a heat energy fraction, consumed for the process of melting of the parent metal, was attained.

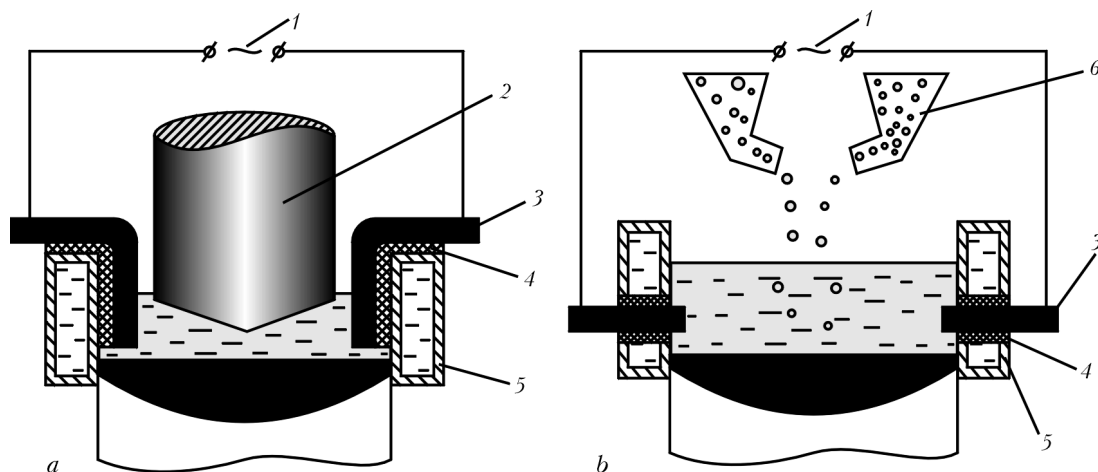


Figure 13 [31]. Scheme of ESR with non-consumable electrodes arranged on the internal wall of the mould (*a*) and through the mould wall (*b*): 1 — power source; 2 — billet being remelted; 3 — non-consumable electrodes; 4 — electric insulators; 5 — mould; 6 — batches with charge materials



This method has found a practical application in processes of electroslag cladding and remelting [34, 35]. Pouring of a molten metal into a mould made it possible to break completely the relation between the remelting current and melting speed of the metal supplied, its temperature and, consequently, between the processes of melting and crystallization of metal. Using this scheme of remelting, the current value is preset coming from required values of slag pool temperature and conditions of crystallization of the ingot being formed. The advantage of the above-mentioned scheme is the independence of rate of molten metal supply to the crystallization front of power at the slag pool. Other advantage is the feasibility of holding of prepared molten metal at the preset temperature, its pouring after cooling to a required temperature, up to the temperature close to liquidus temperature. This guarantees the favorable conditions for initiation of numerous centers of crystallization in a metal pool volume and contributes to the formation of a fine-dispersed disoriented crystalline structure.

The pouring of molten filler metal (separation of process of melting and crystallization of metal) was realized also earlier in the processes of electroslag and portion electroslag casting of large (from 5 up to 200 t) ingots [36, 37]. Here, the current to a slag pool was supplied by non-consumable electrodes (Figure 14). The portion electroslag casting (PESC) is used mainly for the production of large forge ingots as a method of prevention of defects of a liquation and shrinkage origin, whose formation in conventional casting is increased with increase in the ingot mass. As in PESC process the molten metal is supplied by portions, a special attention is paid to the quality of joining (transition zone) between separate portions.

At present the electroslag technologies in a current-carrying mould using a liquid metal (EST LM) envisage also a portion pouring. This is explained by the fact that the existing pouring devices cannot provide the continuous pouring of metal at a required rate and temperature of the metal being poured. Requirements to the devices for EST LM are described

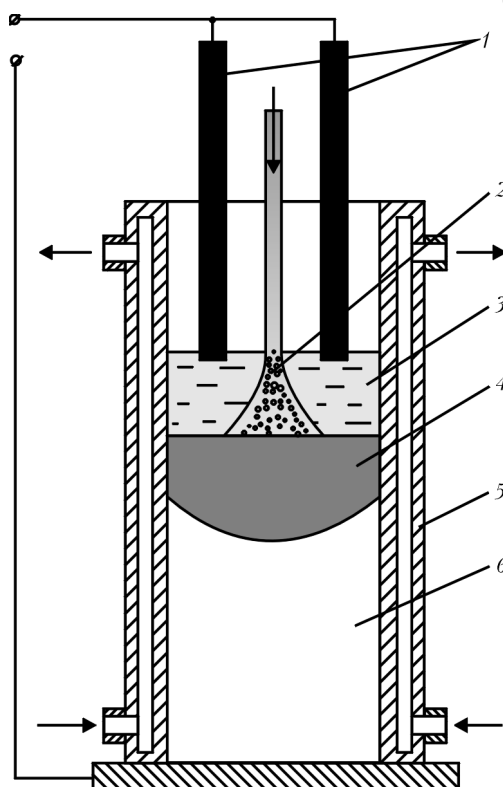


Figure 14. Scheme of electroslag casting of ingots: 1 — non-consumable electrodes; 2 — molten metal jet; 3 — slag pool; 4 — metal pool; 5 — mould; 6 — ingot

in work [38]. The portion pouring of molten metal in cladding into a current-carrying mould can lead not only to the formation of a thin white band in a macrostructure (as also in PESC process), but also to the formation of a wavy zone of joining of parent and clad metals (Figure 15). As was noted in work [39], the identity of penetration profile before and after pouring of a portion of a molten filler metal proves that the penetration profile is defined by a nature of melting of a parent metal in a slag pool. The effect of molten metal supply is negligible. At similar rate of molten metal supply to the front of crystallization (increase in frequency of pouring of separate portions and appropriate decrease in their volume), i.e. approaching the condition of a continu-

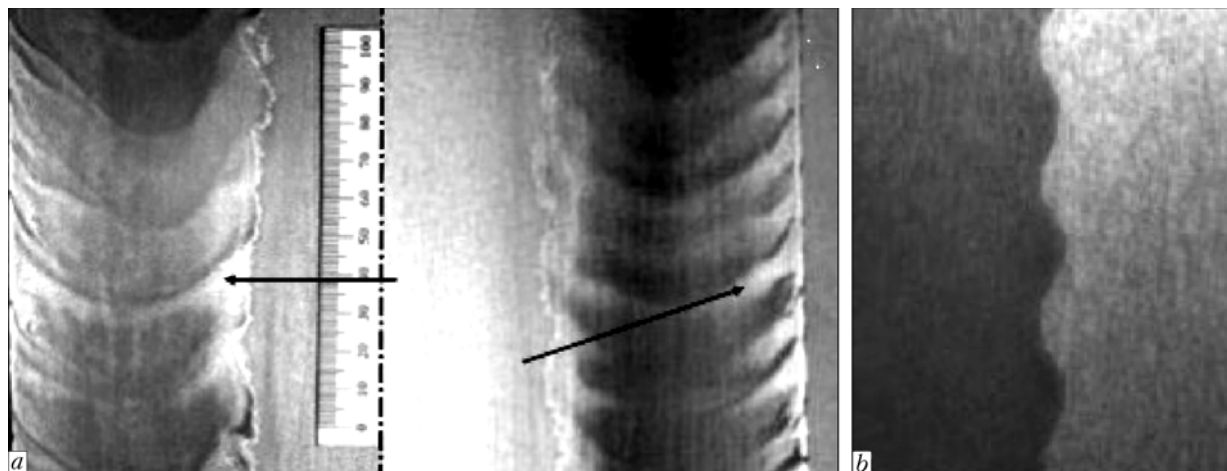


Figure 15 [39]. Fragment of the joint zone in portion cladding with liquid metal in a current-carrying mould: *a* — white bands in a cladding macrostructure; *b* — penetration profile

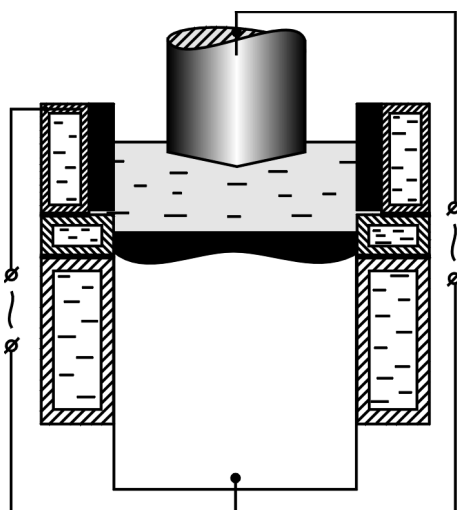


Figure 16. Double-circuit scheme of consumable electrode remelting

ous supply of the molten filler metal, the zones of joining indicated in the Figure are disappeared, it is possible to attain almost smooth line of fusion.

The progress of electroslag technologies in a current-carrying mould promoted the development of a so-called double-circuit diagram of consumable electrode remelting (Figure 16) [40, 41], developed because of absence of capabilities at the present time for holding and supply of liquid metal melt, in particular including easily-oxidizing elements, for example titanium, aluminium, into a current-carrying mould. In the new diagram of remelting of a consumable electrode in a current-carrying mould, as compared with the scheme of Figure 13, the effectiveness of electrode melting is increased owing to such a powerful source of heat generation as a contact resistance between the consumable electrode and slag pool. When necessary, it is possible to increase the speed of melting of the consumable electrode without increase in electric power in the mould circuit. This allows control the shape and volume of the metal pool within the wide ranges, and, thus influencing the susceptibility of metal of cast ingots of high-alloy steels and alloys to the formation of such crystallization defects as a speckle liquation.

Separation of processes of melting and crystallization of metal in a current-carrying mould, appearance of a double-circuit diagram of ESR of consumable electrode open up the new opportunity for the development of electroslag technologies. First of all, it is the development of a process of continuous or semi-continuous electroslag casting of ingots. This process requires a device providing continuous, at comparatively low rates (up to 1000 kg / h) [38, 42], supply of a molten metal of a required temperature into the current-carrying mould. Moreover, the continuous supply of the molten metal will allow elimination of formation of structural changes in the zone of joining of separate portions, in particular in melting of hollow ingots, as well as formation of a wavy zone of joining in cladding.

The feasibility to provide a current supply to a slag pool at any place of the mould, and, consequently, to separate the zones of maximum heat generation facilitates the producing of slab and shaped ingots with the quality surface (i.e. formation of narrow sides and angles of the ingot). It is necessary to define the optimum scheme of current supply for flat and other current-carrying moulds. Thus, it is possible to connect either narrow or wide panels of a current-carrying section, and possibly all together to one of poles of the power source for the flat mould. In making the electroslag process in a flat panel current-carrying mould it is possible to use also the natural vertical slots of the mould. The different schemes of connection of separate sections of the mould and also use of additional current-carrying elements allow influence the hydrodynamic flows in slag and metal pools and, consequently, widening the capabilities in control of the ingot crystallization process.

Arrangement of current-carrying elements on the bottom of a slag pool facilitates the design of a multi-passage current-carrying mould. Producing of a large amount (six, eight or more) of small-section ingots, is, evidently, possible using appropriate pouring devices used in foundry.

The feasibility to keep any (in accordance with properties of a slag used) preset temperature of a slag pool and to provide here the supply at a preset speed of a molten filler metal of a required temperature to the front of crystallization allows realizing this process as cladding without penetration of a cladding material. Rationality and need in this process of cladding-brazing are described in works [42, 43]. Using cladding without penetration, it is possible to produce not only bimetal products with minimum thickness of a clad layer, but also the quality joints of dissimilar metals, the fusion welding of which leads to the formation of brittle intermediate structures. If there is no penetration, no intermetallics are observed in the joint zone. Naturally, a proper slag is required for similar processes [44]. Depending on the purpose, the slags can be both low- and also high-temperature.

Not all the list of trends was given, according to which it is possible to update the electroslag technologies using a current-carrying mould and molten filler metal, however, the solution of the put problems will allow widening the capabilities of a special electrometallurgy.

1. Latash, Yu.V., Matyakh, V.N. (1987) *Current methods of production of super high quality ingots*. Kiev: Naukova Dumka.
2. Paton, B.E., Medovar, B.I., Shevtsov, V.L. et al. (1975) ESR of rectangular section ingots in mould with ribbed working wall. In: *Special electrometallurgy*, Issue 27, 3–8.
3. Paton, B.E., Kuzema, I.D., Medovar, B.I. et al. (1974) On problem of selection of types and sizes, parameters and shape of ESR moulds for melting of large slab ingots. *Ibid.*, Issue 25, 19–28.
4. Paton, B.E., Medovar, B.I., Artamonov, V.L. et al. (1972) Electroslag remelting with ingot drawing from single- and multi-passage moulds. *Ibid.*, Issue 14, 3–12.
5. Grodzinsky, S.V., Baglaj, V.M., Fedorovsky, B.B. et al. (1977) Mould for melting of hollow ingots. *Ibid.*, Issue 34, 55–58.



6. Medovar, B.I., Chekotilo, L.V., Popov, V.G. et al. (1970) Electroslag melting of hollow billets of variable-height section. *Ibid.*, Issue 8, 41–42.
7. Paton, B.E., Medovar, B.I., Chekotilo, L.V. et al. (1971) Electroslag melting of rectangular hollow ingots. *Ibid.*, Issue 13, 35–39.
8. Timashov, G.A., Fedorovsky, B.B., Krepak, V.A. et al. (1984) Experience of introduction of electroslag piercing in production of shaped hollow billets. *Ibid.*, Issue 54, 44–46.
9. Fedorovsky, B.B., Timashov, G.A., Bondarenko, L.I. et al. (1986) ESC of long billets having simultaneously some holes of small diameters. *Problemy Spets. Elektrometallurgii*, 3, 38–39.
10. Paton, B.E., Medovar, B.I., Kumysh, I.I. et al. (1976) Availability of production of kiln bands by ESC method. In: *Special electrometallurgy*, Issue 32, 3–6.
11. Tsykulenko, K.A., Fedorovsky, B.B., Tsykulenko, A.K. (2005) Application of bimetal and refractory materials in ESR mould designs. *Advances in Electrometallurgy*, 1, 14–16.
12. Medovar, B.I., Bogachenko, A.G., Artamonov, V.L. et al. (1973) New panel structures of assembled ESR moulds. In: *Special electrometallurgy*, Issue 21, 38–42.
13. Medovar, B.I., Shevtsov, V.L., Marinsky, G.S. et al. (1977) Study of temperature deformations of ESR mould panels. *Ibid.*, Issue 33, 60–63.
14. Dovzhenko, A.F., Moderau, Yu.V., Berman, K.A. et al. (1983) Mould: USSR author's cert. 612511. Int. Cl. C21C 5/56. *Ibid.*, Issue 53, 84.
15. Kiselman, M.A., Tsikerman, A.L., Volokhonsky, L.A. et al. (1983) Mould: USSR author's cert. 6626588. Int. Cl. C21C 5/56. *Ibid.*, Issue 53, 86.
16. Medovar, B.I., Makhnenko, V.I., Olejnik, V.A. et al. (1983) Influence of cooling conditions on stress state of hollow ESR ingots. *Ibid.*, Issue 59, 16–24.
17. Ilchenko, A.I., Kuchin, G.P., Marinsky, G.S. et al. (1973) Melting of electroslag ingots in moulds cooled by boiling layer of intermediate heat medium. *Ibid.*, Issue 22, 38–42.
18. (1978) Method and unit for ESR: Pat. 287215 Austria. Int. Cl. B22D 9/00, 31B 3/03. *Ibid.*, Issue 36, 10.
19. (1978) Mould for remelting of metals with protecting slag: Pat. 2081724 France. Int. Cl. B22D 11/00, C21C 7/00, F27D 1/00, 11/00. *Ibid.*, Issue 36, 95.
20. Medovar, B.I., Fedorovsky, B.B., Tsykulenko, K.A. et al. (1989) Decrease of heat losses from slag pool in ESR. *Problemy Spets. Elektrometallurgii*, 4, 22–25.
21. Tuchkevich, N.M., Grechin, V.P., Klyuev, M.M. et al. (1973) Influence of reverse rotation of mould on ESR process and steel quality. In: *Special electrometallurgy*, Issue 19, 34–44.
22. Medovar, B.I., Stupak, L.M., Mokhnach, V.K. et al. (1970) On problem of electromagnetic action on solidification of ESR ingot. *Ibid.*, Issue 4, 16–20.
23. (1978) ESR method of consumable electrodes: Pat. 2047202 Germany. Int. Cl. C21C 5/52, 18B 5/52. *Ibid.*, Issue 36, 80.
24. Kazimirov, A.N., Mukhin, Yu.M., Gonchar, Ya.S. et al. (1981) Quality of metal of horizontal ingots melted in tilt mould during ESR. *Ibid.*, Issue 45, 20–22.
25. Medovar, B.I., Tsykulenko, A.K., Skripnik, V.P. et al. (1981) Examination of dissolution process of lump filler materials in horizontal ingot metal and their modifying action on structure and quality of ingots. *Ibid.*, Issue 45, 23–31.
26. Kolisnyk, V.N. (1970) New scheme of electroslag melting. *Ibid.*, Issue 5, 8–13.
27. Kolisnyk, V.N. (1970) About new possibilities of electroslag melting in production of metals from chemical compounds. *Ibid.*, Issue 7, 17–21.
28. Volkotrub, N.P., Latash, Yu.V. (1981) Examination of possibility of metallic component extraction from slime by electroslag melting. *Ibid.*, Issue 46, 32–36.
29. Ksyondzyk, G.V. (1975) Current-carrying mould providing the rotation of slag pool. *Ibid.*, Issue 27, 32–40.
30. Rabinovich, V.I., Kriger, Yu.N., Svitenko, I.A. et al. (1982) Unit for electroslag melting: USSR author's cert. 418093. Int. Cl. C21C 5/56. *Ibid.*, Issue 48, 128.
31. Mkhitarian, G.A., Shakhverdyan, S.L. (1981) Electroslag furnace with dispersed heat generation in slag pool. *Ibid.*, Issue 47, 51–54.
32. Nikitin, V.P. (1961) *Selected transactions*. Moscow: AN SSSR, 109–112.
33. (1947) Method of welding of metals with separate melting processes. *Doklady AN SSSR*, 56(5), 477–480.
34. Medovar, B.I., Tsykulenko, A.K., Medovar, L.B. et al. (1997) Electroslag processes without consumable electrodes. *Problemy Spets. Elektrometallurgii*, 2, 12–16.
35. Medovar, B.I., Chernets, A.V., Medovar, L.B. et al. (1995) Electroslag cladding with liquid filler metal. *Ibid.*, 1, 6–11.
36. Latash, Yu.V., Voronin, A.E., Nikolaev, V.A. (1975) Production of high-quality large ingots by portion electroslag casting method. *Stal*, 11, 999–1002.
37. Latash, Yu.V., Voronin, A.E., Nikolaev, V.A. et al. (1975) New method of production of high-quality large forging ingots. In: *Special electrometallurgy*, Issue 2, 31–42.
38. Tsykulenko, K.A. (2002) Unit for pouring of liquid metal. Requirements for these units of EST LM. *Ibid.*, 2, 12–14.
39. Tsykulenko, K.A. (2007) Certain aspects of deposited layer formation in portioned electroslag cladding with application of liquid additive metal. *Advances in Electrometallurgy*, 2, 6–9.
40. Tsykulenko, A.K., Lantsman, I.A., Medovar, L.B. et al. (2000) Double-circuit of electroslag remelting of consumable electrode. *Problemy Spets. Elektrometallurgii*, 3, 16–20.
41. Medovar, B.I., Medovar, L.B., Tsykulenko, A.K. et al. (1999) On problem of electroslag melting of large-tonnage billets from high-alloy steels and alloys. *Ibid.*, 2, 26–30.
42. Tsykulenko, K.A. (2002) Electroslag cladding with liquid metal. In: *Electroslag Cladding with Liquid Metal: Yesterday, Today, Tomorrow*: Proc. of Int. Sci.-Techn. Conf. Kiev, 118–122.
43. Tsykulenko, K.A., Tsykulenko, A.K. (2001) Analysis of feasibility of electroslag cladding of dissimilar metals and alloys at small thickness of deposited layer. *Problemy Spets. Elektrometallurgii*, 3, 19–23.
44. Tsykulenko, K.A., Tsykulenko, A.K. (2001) To problem of flux selection for electroslag cladding–brazing. *Ibid.*, 4, 3–6.



EFFECT OF SUBSTRATE TEMPERATURE ON MICRO- AND SUBSTRUCTURE OF COPPER CONDENSATES DEPOSITED FROM A VAPOR PHASE

A.I. USTINOV, E.V. FESYUN, T.V. MELNICHENKO and S.M. ROMANENKO

E.O. Paton Electric Welding Institute, NASU, Kiev, Ukraine

Effect of temperature of substrate on micro- and substructure of copper condensates, formed from a vapor phase by the method of electron beam evaporation, was investigated. It was established that the decrease in temperature of the substrate leads not only to the decrease in sizes of grains of the copper condensates, but also to the qualitative change in their substructure, i.e. nanotwins are formed by the system $\{111\}$, $\langle 112 \rangle$. The change in substructure of copper condensates is accompanied by an abrupt increase in their microhardness.

Keywords: electron beam evaporation and condensation, microhardness, transmission electron microscopy, twins

Introduction. Microstructure of condensates depends on the conditions of deposition of a vapor phase. The most important factor influencing the characteristics of the microstructure (mainly size and shape of grains, and also degree of condensate consolidation) is the substrate temperature [1, 2] at the reduction of which the sizes of microstructural elements of vacuum condensates are decreased. In accordance with the Hall-Petch relation, these changes of the microstructures of condensates should be accompanied by an increase in material yield strength. As the theoretical analysis shows it is possible to expect the large changes in mechanical properties of condensates with decrease in sizes of microstructural elements to a nanoscale level [3]. Therefore, a great attention is paid recently to the study of conditions of producing condensates with a nano-dimensional scale of structural elements.

At the same time, it is outlined in some publications that with reduction in the substrate temperature not only the grain size is decreased, but also the density of defects of a crystalline structure, i.e. defects of a vacancy type (dislocation loops, micropores, etc.), dislocations and planar defects (packing defects, twins) is increased. Thus, for example, in case of copper condensates the increase in density of dislocations was observed with a reduction of the substrate temperature and some technological parameters of deposition (film thickness, pressure of residual gases in a chamber, etc.) [4, 5]. Formation of planar defects (packing defects, twins) was also observed in condensates on the base of metals with a face-centered cubic lattice [6]. Consequently, the change in conditions of formation of condensates can be accompanied by the change not only in sizes of microstructural elements, but also in characteristics of their substructure. It is necessary to take this into account in the development of a vapor-phase technology of creating

of materials and coatings characterized by high values of mechanical characteristics.

On the example of copper condensates, produced in electron beam deposition from a vapor phase in vacuum, an effect of substrate temperature on characteristics of their micro- and substructure was studied. On the basis of the obtained results the comparison of mechanical properties (microhardness) of copper condensates and characteristics of their micro- and substructure was made.

Procedure of producing samples and conductance of experiment. The copper condensates of up to 100 μm thickness were produced by the method of electron beam evaporation of 50 mm diameter copper ingot in vacuum and deposition of a vapor metal flow on the substrate. The temperature gradient was created along the substrate that allowed producing condensates under similar conditions at different temperatures of the substrate. Using this method, the condensates in the range of the substrate temperature from 165 up to 700 $^{\circ}\text{C}$ were produced. The deposition rate of condensates in this case was 2–3 $\mu\text{m} / \text{min}$. To separate condensates from the substrate, a layer of salt (NaCl or CaF_2) was placed preliminary on it.

Microstructures of condensates and their chemical composition were examined on transverse sections using scanning electron microscope CamScan 4, equipped with a local chemical analysis system ENERGY 200. The transverse sections were prepared by a standard procedure with a subsequent chemical etching of a polished surface. Substructure of condensates was examined on thin foil samples prepared by thinning of the condensate foil by a jet electrolytic polishing (chemical composition of electrolyte: 45 ml of nitric acid and 255 ml of water at voltage of 15–20 V and room temperature of electrolyte) using a transmission electron microscope HITACHI-800 at 200 kV accelerating voltage.

X-ray examinations (texture analysis) were made on separate condensates in K_{α} -radiation of a copper

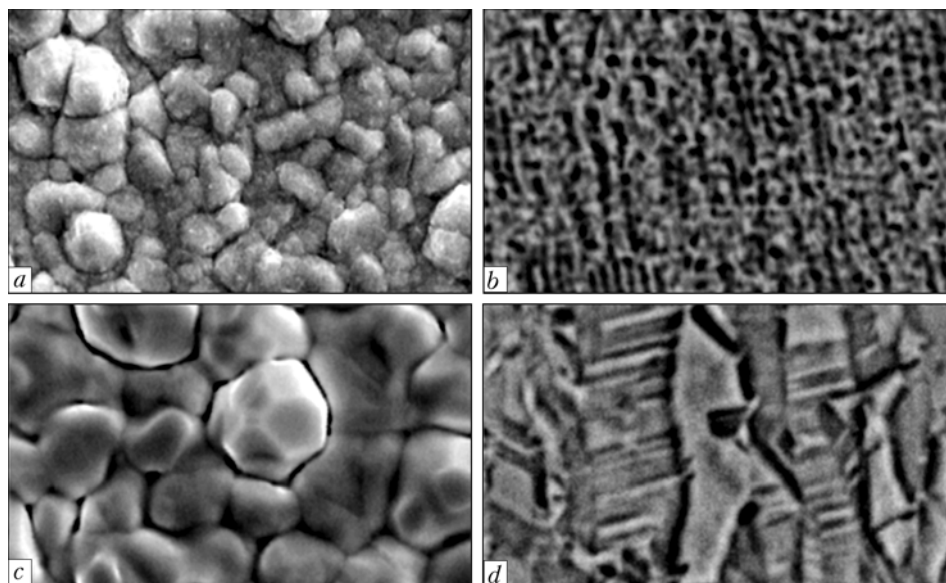


Figure 1. Microstructure ($\times 300$) of surface and cross section of condensates deposited at temperature values 165 (a, b) and 540 (c, d) °C

anode in diffractometer DRON-3, equipped with a texture attachment, allowing measuring in an automatic condition the intensity of a dissipated radiation at different orientations of a foil plane relative to a primary beam. From the data obtained, stereographic projections of pole densities were plotted for different crystallographic planes of copper.

Microhardness by Vickers method was determined on transverse sections using an optic microscope Polyvar Met, equipped with an attachment for measurement of hardness at 5 g load to indenter and 10 s loading duration.

Experimental data. Condensates of copper had purity, defined mainly by a purity of initial ingot of not worse than 99.8 wt.% (main impurities of ingot are such metals as 0.05 Al, 0.02 Ni and 0.03 Zn).

With varying of the substrate temperature the change in a surface relief of condensates and their microstructure was observed. Figure 1 presents examples of a typical type of the surface relief and microstructure of transverse sections of copper condensates deposited at different values of the substrate temperature: at 165–350 °C the condensates are formed whose

surface relief has a dome-like shape (on transverse sections of these condensates a mechanical texture is seen, the crystallites of which have a shape, elongated along the direction of growth, the width of their cross section is less than 1 μm ; Figure 1, b); above 350 °C the condensates are formed, whose surface relief is characterized by the presence of clearly cut crystallites (Figure 1, c). Columnar shape of crystallites is preserved, and their width is increased.

Distribution of cross section of crystallites by sizes for condensates, produced at different substrate temperatures, is given in Figure 2. As follows from the Figure, the distribution of crystallites by size D can be described by a symmetric curve with a peak, whose position is shifted to the region of smaller sizes with decrease in the substrate temperature.

Figure 3 shows the change in width of columnar crystallites in copper condensates depending on the substrate temperature. It should be noted that with reduction of the substrate temperature, the width of columnar crystallites of copper condensates is decreased by a linear law.

X-ray examinations of copper condensates, produced at different substrate temperatures, showed

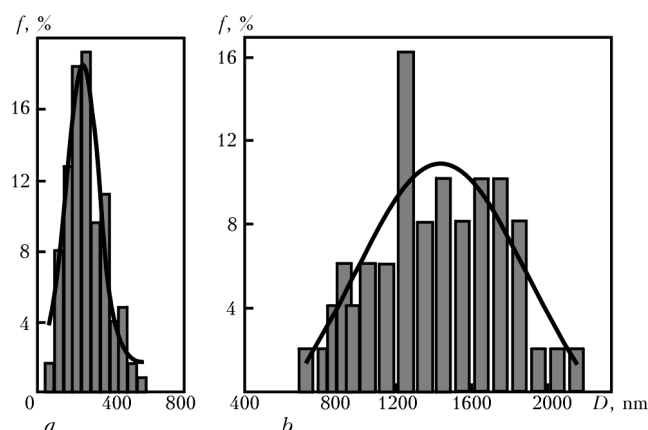


Figure 2. Distribution of columnar crystallites by sizes at temperature values 165 (a) and 320 (b) °C: f — share of columnar crystallites of a definite size

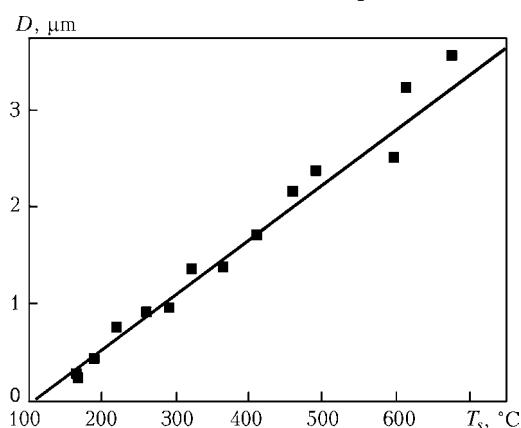


Figure 3. Dependence of width D of columnar crystallites of copper condensates on substrate temperature T_s

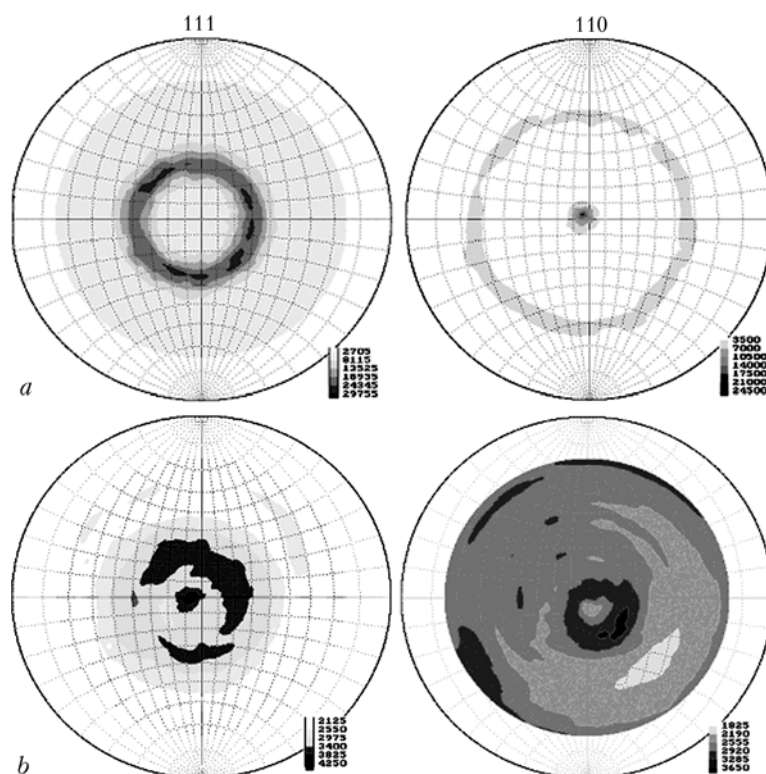


Figure 4. Pole figures of distribution of pole densities 111 and 110 for condensates deposited at temperature values 700 (a) and 300 (b) °C

that the presence of crystallographic texture is also characteristic of it. It was established that with reduction in the substrate temperature the type of crystallographic texture of condensates is changed in the region of values of the substrate temperature of 350 °C.

Figure 4 gives typical examples of distribution of pole densities (111) and (110) for condensates deposited at different temperatures. At increased temperatures of the substrate (above 350 °C) the condensates are formed with a single-component texture $\langle 110 \rangle$ of a fiber type, that is proved by a peak of pole density 110 in the center of pole figures and peak of pole density 111, located at the distance of 35 angular degrees. With decrease in the substrate temperature the component of a fiber texture $\langle 110 \rangle$ is weakened: instead of peak of pole density in the center of a pole figure a ring of pole density, located at the distance of 19 angular degrees from a center of pole figure, and peak of pole density for planes 111, located in the center of pole figure, are formed. These changes prove the appearance of a two-component texture $\langle 111 \rangle + \langle 112 \rangle$.

Thus, it follows from the conducted texture analysis of condensates deposited at different values of the substrate temperature that with decrease in temperature the orientations of crystals along the direction $\langle 111 \rangle$ and $\langle 112 \rangle$ become most probable, while for condensates, deposited at temperature above 350 °C, the crystallites are oriented along the direction $\langle 110 \rangle$.

Electron microscopic representations of the microstructure of foil prepared by thinning of condensates, deposited at different temperatures of substrate, are given in Figure 5. It was determined that at increased temperatures of substrate (above 350 °C) the conden-

sates of copper with coarse grains are formed, whose size can reach several micron (Figure 5, a). Internal structure of grains is characterized by the presence of defects of a dislocation type. When depositing condensate on substrates, located at the lower temperatures (below 350 °C), the grain size is decreased to a submicron scale, and separate grains are divided into domains with straight-linear boundaries, which, as a rule, are crossing the entire grain (Figure 5, b). With decrease in temperature of substrate (Figure 5, c), the number of grains, containing these domains, is increased. It should be noted that the boundaries between domains in each separate grain are parallel each other, i.e. only one orientation of domains is observed in one grain.

Similar substructure of grains are typical of crystals containing twins [6, 7]. As microdiffraction pattern, taken from the region of crystal contained these domains, proves, the fragments of grains divided by straight-linear boundaries, are located in a twin orientation with respect to each other in plane (111), i.e. boundaries of domains are the boundaries of twins by system (111), [112].

It is possible to reveal a polydomain substructure of crystallites on transverse sections of condensates at a high magnification: boundaries of domains are located both at angle 20–60°, and also almost normal to the direction of growth of crystallites (Figure 6).

Let us analyze the thickness of twin domains for the quantitative evaluation of twin boundaries in condensates. Figure 7 shows examples of distribution of twin domains in thickness for copper condensates deposited at different temperatures of substrate. It is evident that the distribution of twin domains in thickness can be described by an asymmetric curve, similar

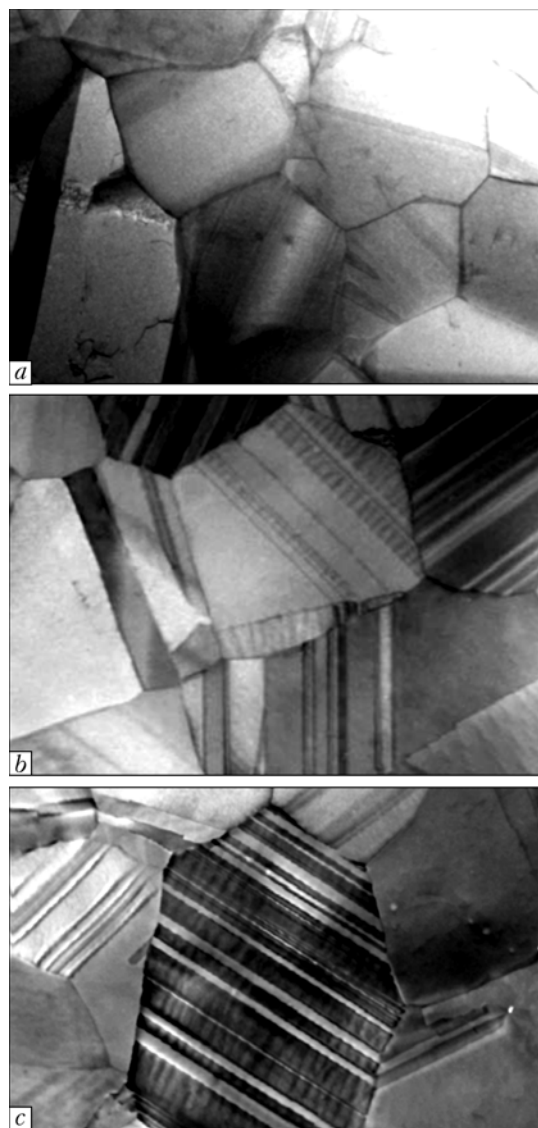


Figure 5. Substructure of crystallites of copper condensates deposited at substrate temperature values 600 (*a* — $\times 10000$), 300 (*b* — $\times 30000$) and 165 (*c* — $\times 40000$) °C

to lognormal distribution, whose position of peak is shifted to the region of higher values of thicknesses of twins with increase in the substrate temperature. Here, the dispersion of distribution of thicknesses of

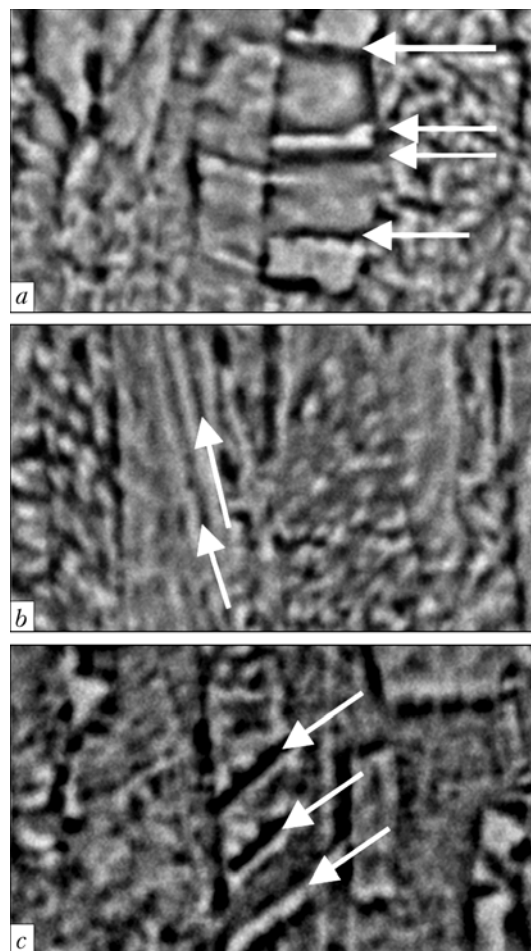


Figure 6. Microstructure ($\times 3000$) of cross sections of copper condensates deposited at substrate temperature 430–600 °C. Arrows show the direction of twin boundaries relative to the direction of growth of crystallites: *a* — 90; *b* — 20; *c* — 45°

twins by sizes is also increased with a temperature growth.

Figure 8 presents dependence of change in mean thickness of twins on substrate temperature. It follows from this Figure that with decrease in substrate temperature to lower than 350 °C, the thickness of twin domains is abruptly decreased and reaches a nano-dimensional scale.

If to assume that the alternating of twin domains *P* with different thicknesses has a random nature (prob-

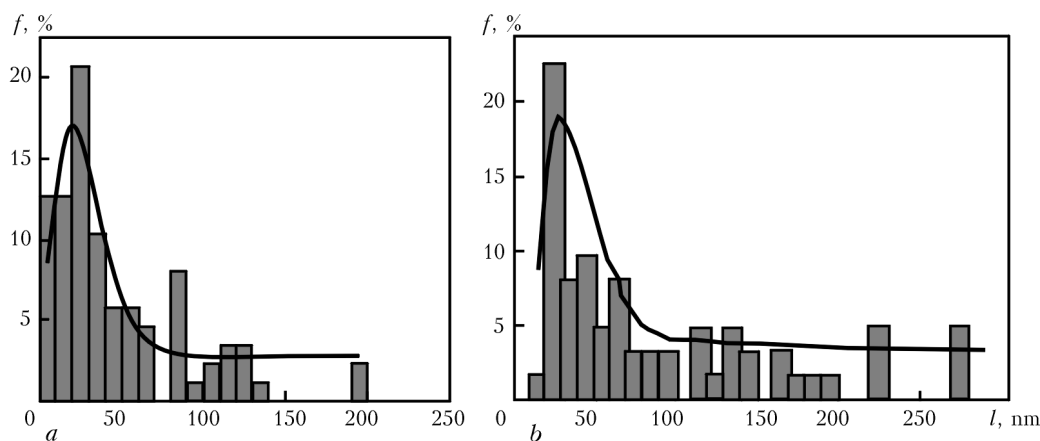


Figure 7. Distribution of twin domains in thickness *l* of copper condensates deposited at the substrate temperature values 165 (*a*) and 320 (*b*) °C: *f* — share of twin domains of a definite size

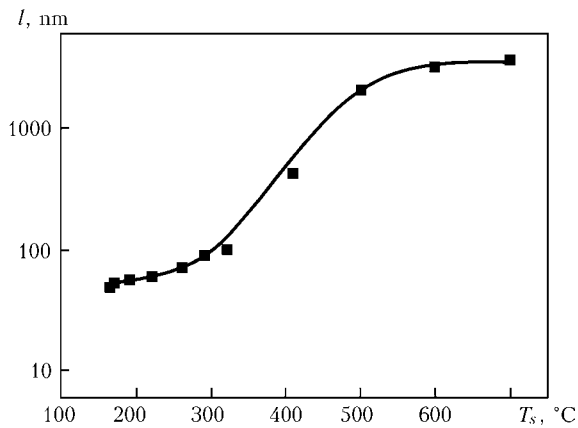


Figure 8. Dependence of a mean thickness of twin domains on substrate temperature

ability of occurrence of a twin boundary at a definite distance from the nearest one does not depend on this distance), then the density of twin boundaries in crystal (ratio of quantity of planes of atoms 111 in the crystal, parallel to plane of twinning, to the quantity of planes 111, being the boundaries of twins) expressed in percents, will be associated with a mean thickness of twins (distance between boundaries of twins) by relation $P = d/l \cdot 100\%$, where d is the interplane distance in the direction normal to plane of twinning 111, nm; l is the thickness of twin domains, nm.

Figure 9 shows change of this value depending on the substrate temperature. With decrease in temperature to 350 °C, the density of twins remains rather low: grains do not almost contain twin boundaries. With a further decrease of the substrate temperature the density of twins begins to increase abruptly. This change in density of twins with reduction in the substrate temperature can be considered as threshold, i.e. there is a critical temperature of deposition, near which a qualitative change of substructure of copper grains, transition of grains from mono- into polydomain state, was observed.

Results of durometric examinations of copper condensates deposited at different value of the substrate temperature are given in Figure 9. With decrease in the substrate temperature the microhardness of condensates is non-monotonously changed. Temperature reduction to 400 °C is accompanied by a negligible

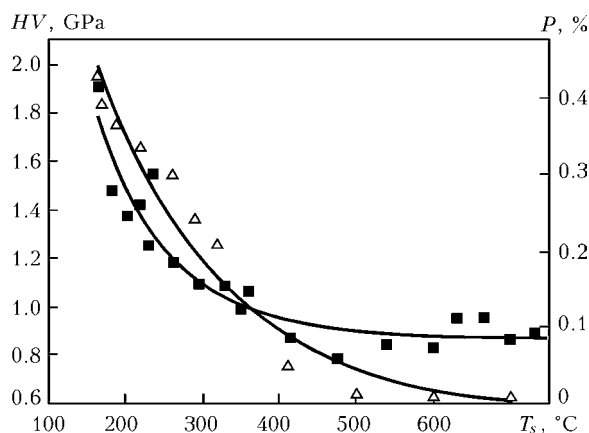


Figure 9. Dependence of density of twin boundaries P (Δ) and microhardness ($\blacksquare$) of copper condensates on substrate temperature

decrease in microhardness of condensates. The further drop in the substrate temperature leads to an abrupt increase in microhardness with decrease in temperature from 350 to 160 °C. It is seen from the comparison of nature of changing the density of twin boundaries and level of microhardness that the change in these characteristics is correlated with each other, i.e. with increase in density of twin boundaries the microhardness of condensates is increased.

Thus, the carried out investigations of micro- and substructure of copper condensates, deposited at different temperatures, showed that such characteristics of the microstructure as size of crystallites, are changed continuously, while the texture and density of twins undergo jumpy changes at temperatures of the substrate below 350 °C. Moreover, the increase in density of twins is accompanied by an abrupt increase in microhardness of condensates.

Discussion of results. Copper with a nanotwin substructure of grains was produced earlier by electric deposition [8–14] and magnetron sputtering [15, 16]. It was assumed in work [10] that such substructure of grains is formed during the process of their growth. Analysis of possible mechanisms of formation of twin substructure in the process of electric deposition of copper condensates showed that the twins can be the result either of formation of a polydomain nuclei or errors of growth of crystallites. Nucleation of a polydomain nuclei can occur in the process of coalescence of inlands, and the violation of a proper sequence of packing of compactly-packed atomic layers, leading to nucleation of twin interlayers is due to the accumulation of impurities at the face of the crystallite growth [10]. In the first case, the boundaries of twins will be parallel to the direction of growth of crystallite, while in the second case the boundaries are located under the angle towards the direction of the crystallite growth.

As the experimental investigations of substructure of grains of copper condensates formed at different substrate temperature, showed, the twin boundaries can be oriented in different ways relative to the direction of growth of the crystallites. In accordance with data of texture analysis, the dominating direction of crystallites at substrate temperature below 350 °C is $\langle 111 \rangle$ and $\langle 112 \rangle$, and at temperatures above 350 °C it is $\langle 110 \rangle$. From the crystallographic point of view, the twin boundaries can be of two types: by planes 111 and 111 for crystallites, oriented to the direction $[111]$ (Figure 7). Twin boundaries of the first type are parallel to the condensate surface and can be revealed only on transverse sections of condensates. Twin boundaries of the second type will cross the foil surface under angle of about 20° and can be observed in electron-microscopic images of microstructure at inclination of the foil surface relative to electron beam.

In case of crystallites oriented to the direction $[112]$, the twin boundary is either parallel to this direction (plane 111) or makes angle with it of approximately 28° (plane of 111 type), or located under the angle of about 70°. Therefore, in these crystallites the twin boundaries of the second and third types will

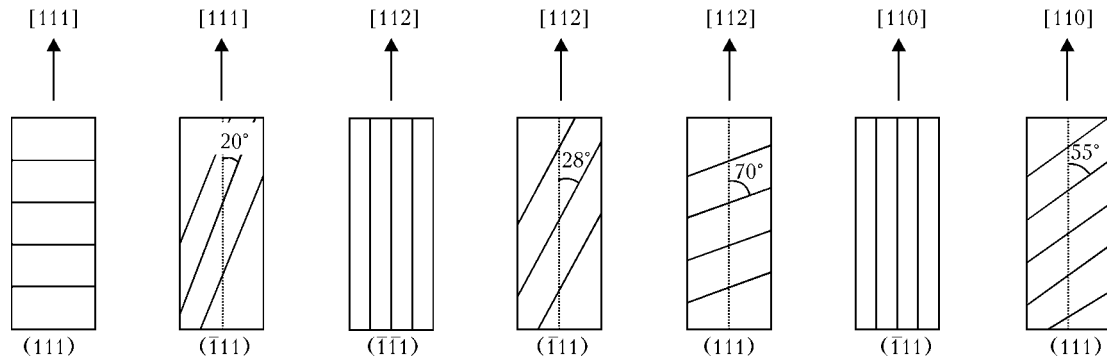


Figure 10. Scheme of possible orientations of boundaries of twins of 111 type with respect to the direction of growth of crystallites

not be revealed in electron-microscopic images of the microstructure of foils prepared parallel to the condensate surface (Figure 10).

In crystallite orientation to the direction [110] the twin boundaries can locate either parallel to the direction of the crystallite growth or under the angle of approximately 55° (see Figure 10). In electron-microscopic examinations of the foil, prepared parallel to the condensate surface, only twin boundaries of the first type can be observed in microstructural images. Twin boundaries of the second type can be observed only on transverse sections of condensates.

It follows from the crystallographic analysis that the non-diffuse twin boundaries can be revealed on electron-microscopic images of microstructure of foils, not using inclination of foils relative to a beam, only in crystallites oriented to the direction 112 and 110, and at high inclinations of the foil — to the direction 111 and 112. Twin boundaries on transverse sections can be observed at any orientation of the crystallites.

It is seen on electron-microscopic images of microstructure of thin foils of condensates, given in Figure 5, that all grains can be divided conditionally into those containing and non-containing the twin boundaries. Crystallographic analysis of all possible orientations of twin boundaries of type 111 and texture analysis of dominating orientation of crystallites allow making conclusion that twin boundaries can be oriented parallel to the direction of growth of crystallites and can locate relative to the crystallite axis under some angle (from 20° to 90°).

As the process of formation of condensates from a vapor phase, the same as in electric deposition, is realized mainly by joining of separate atoms, then mechanism of formation of a twin structure in case of vapor condensation can be assumed similar to that suggested in work [10] for the case of electric deposition. However, unlike the process of formation of condensates in electric deposition, the temperature of substrate in condensation of a vapor flow influences greatly the substructure of condensates. This can be stipulated by two factors. First, the substrate temperature has a great influence on diffusivity of atoms on the surface of crystallites, and, secondly, in varying the substrate temperature the ratio between rates of formation of nuclei of crystallites and their growth is changed. Taking this into consideration, the probability of dominating of the processes, caused by a diffusivity of atoms, can be assumed with increase in the substrate temperature, that promotes the appear-

ance of segregations of insoluble impurities at the faces of growing crystals and will stimulate their growth. With temperature decrease the nucleation of a large amount of nuclei will predetermine their substructure. Therefore, at low values of temperature there appears the great probability of formation of polydomain nuclei, while at increased temperatures there appears probability of nucleation of twin domains at the faces of growing crystallites due to formation of segregations of insoluble impurities. Consequence of realization of these processes is, probably, the absence of twin boundaries in microstructure images of foils, produced by thinning of high-temperature condensates, and their presence on transverse sections, whereas in case of low-temperature condensates the twin boundaries are observed mainly on foils.

To clarify the role of a twin substructure of grains in a mechanical behavior of condensates, the dependence of microhardness on a mean size of twin domains ($d_1 = 1/2 (l + D)$, where l is the thickness of twin interlayer; D is the length of twin interlayer on sections parallel to the condensate surface) was plotted in coordinates of Hall–Petch.

It is seen from Figure 11 that the dependence of microhardness on a mean size of domains is described by two straight-linear areas with characteristic inclinations. The first region corresponds to condensates, for which the size of twin interlayers is commensurable with a cross section of columnar crystallites, while the second region corresponds to condensates, for which the thickness of twin interlayers becomes much smaller than a cross section of crystallites. Mechanical behavior of condensates in both case is defined by dislocation processes (linear dependence of Hall–Petch is fulfilled). At the same time, it can be seen by comparing the angles of inclination at different regions of the Hall–Petch dependence that at the second region (corresponding to a nanotwin substructure of grains) the inclination of a linear dependence is larger than for condensates with microscale sizes of the substructure elements. Increase in angle of inclination at the region, corresponding to nanotwin substructure of condensates can be considered as a prove that the boundaries of twin domains are the more effective barriers of displacement of dislocations than the boundaries of grains.

The effect of twin substructure of grains on mechanical behavior of copper was investigated on condensates produced by a method of electric deposition and magnetron sputtering [8, 9, 11, 13, 17]. Thus, in

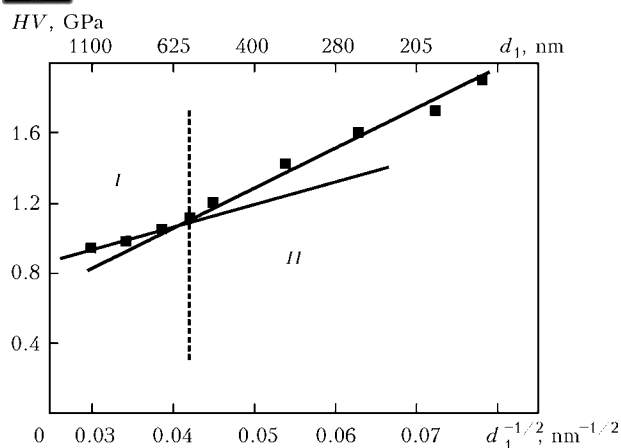


Figure 11. Dependence of Hall-Petch for copper condensates; region I corresponds to condensates with a mono-domain substructure of grains (thickness of twin domains is commensurable with thickness of crystallites); region II — with a nanotwin substructure of grains

work [11] the copper condensates with a similar size of grain (about 400–500 nm) of 99.998 at.% purity, but with a different density of twins, produced by the method of a pulsed electric deposition demonstrated at their nano-identification the different level of hardness: copper samples with a high density of twins had the microhardness at the level of approximately 2.0–2.6 GPa, with a mean density of twins — 1.7–2.0 GPa, and with a low density of twins — 1.4–1.6 GPa. In examination of structure of samples after nano-identification and tensile tests a high density of dislocations was revealed in twin domains and at the boundaries of twins, as well as the shifting of coherent boundaries of twins. The authors made conclusion on this basis that the coherent boundaries of twins are a barrier for the dislocation motion.

A satisfactory conformity to the Hall-Petch law was obtained for condensates with a twin substructure of grains, formed in magnetron deposition of condensates [5].

Thus, it can be concluded from comparing the results of experimental and theoretical investigations of condensates, produced by electric deposition or magnetron sputtering that the nanotwin substructure of copper grains is the determinative factor providing a high level of hardness and strength.

CONCLUSIONS

1. It was established that the consolidated copper condensates, produced by electron beam deposition in vacuum on substrates at 165–700 °C, can be divided conditionally by characteristics of their substructure into high- (substrate temperature is above 350 °C) and low-temperature condensates (substrate temperature is below 350 °C): the first ones are characterized mainly by a mono-domain substructure of grains, while the latter are characterized by a poly-domain substructure. Moreover, the polydomain substructure of each grain is formed by twin domains of a similar orientation, whose thickness has a nano-dimensional scale. Transition from the substructure of high-temperature condensates to the substructure of low-temperature condensates was observed within the

narrow interval of change in the substrate temperature and accompanied by change in type of the crystallographic texture.

2. It was determined that the microstructure of copper condensates, produced in the 165–700 °C range of temperatures of substrate, is characterized by columnar crystallites, which cross-section size is decreased continuously with decrease in the substrate temperature.

3. It was shown that the transition from the substructure of high-temperature condensates to the substructure of low-temperature condensates is accompanied by an abrupt increase in their microhardness. It was concluded from the dependencies of Hall-Petch, plotted for high- and low-temperature condensates in microhardness–mean size of domain coordinates, that the boundaries of twins are the more effective barriers preventing the motion of dislocations that the grain boundaries.

The authors are thankful to A.A. Nekrasov for assistance in conductance of texture investigations of copper condensates. The work has been fulfilled within the scopes of an integrated scientific-technical program of the NAS of Ukraine «Nanosystems. Nanotechnologies. Nanomaterials» (Project No.16/06-N).

1. Movchan, B.A., Demchishin, A.V. (1969) Examination of structure and properties of thick vacuum condensates of nickel, titanium, tungsten and zirconium dioxide. *Fizika Metallov i Metallovedenie*, 28(4), 653–660.
2. Thornton, J.A. (1977) High rate thick film growth. *Ann. Rev. Mater. Sci.*, 7, 239.
3. Andrievsky, R.A., Glezer, A.M. (1999) Dimensional effects in nanocrystalline materials. Specifics of structure. Thermodynamics. Phase equilibrium. Kinetic phenomena. *Fizika Metallov i Metallovedenie*, 88(1), 50–73.
4. Palatnik, L.S., Fuks, M.Ya., Iliinsky, A.I. et al. (1966) Structure and mechanical properties of vacuum copper condensates. *Ibid.*, 22(5), 744–751.
5. Palatnik, L.S., Fedorov, G.V., Iliinsky, A.I. (1961) Substructure and microhardness of vacuum copper condensates. *Ibid.*, 11(5), 815–816.
6. Pashley, D.W. (1965) The nucleation, growth, structure and epitaxy of thin surface films. *Advances in Physics*, 14, 327.
7. Hirsh, P., Hovi, A., Nickolson, P. et al. (1968) *Electron microscopy of thin crystals*. Moscow: Mir.
8. Lu, L., Shen, Y.F., Chen, X. et al. (2004) Ultrahigh strength and high electrical conductivity in copper. *Science*, Vol. 304, 422.
9. Ma, E., Wang, Y.M. (2004) Strain hardening and large tensile elongation in ultrahigh-strength nano-twinned copper. *Appl. Phys. Lett.*, 85(21), 4932.
10. Wang, J., Tian, M., Mallouk, Th.E. et al. (2004) Microtwinning in template-synthesized single-crystal metal nanowires. *J. Phys. Chem. B.*, 108, 841–845.
11. Lu, L., Schwaiger, R., Shan, Z.W. et al. (2005) Nano-sized twins induce high rate sensitivity of flow stress in pure copper. *Acta Materialia*, 53, 2169–2179.
12. Shen, Y.F., Lu, L., Lu, Q.H. et al. (2005) Tensile properties of copper with nano-scale twins. *Scripta Materialia*, 52, 989–994.
13. Shen, Y.F., Lu, L., Dao, M. et al. (2006) Strain rate sensitivity of Cu with nanoscale twins. *Ibid.*, 55, 319–322.
14. Dao, M., Lu, L., Shen, Y.F. et al. (2006) Strength, strain-rate sensitivity and ductility of copper with nano-scale twins. *Acta Materialia*, 54, 5421–5432.
15. Chen, J., Lu, L., Lu, K. (2006) Hardness and strain rate sensitivity of nanocrystalline Cu. *Scripta Materialia*, 54, 1913–1918.
16. Zhang, X., Miera, A., Hoagland, R.G. (2006) Synthesis, microstructure and properties of metallic materials with nanoscale growth twins. In: *Proc. of 25th Army Sci. Conf.*, 18–20.
17. Meyers, M.A., Mishra, A., Benson, D.J. (2006) Mechanical properties of nanocrystalline materials. *Progress in Materials Sci.*, 51, 427–556.



MEASUREMENT OF TEMPERATURE OF DRUM-COOLER IN THE PROCESS OF SUPER-RAPID HARDENING OF METAL MELTED IN PLASMA FURNACE

V.A. SHAPOVALOV, Yu.A. NIKITENKO, V.R. BURNASHEV and G.F. TORKHOV

E.O. Paton Electric Welding Institute, NASU, Kiev, Ukraine

Problems of calorimetry of a rotating drum-cooler during super-rapid hardening of the melt are considered. Optimum scheme of measurement was selected. Investigations were carried out allowing determination of heat transfer between the melt and drum-cooler.

Key words: spinning, thermoresistor, specific resistance, absorbed heat

In production of metals with amorphous and microcrystalline structure the methods of a super-rapid hardening from a molten state (extraction, spinning) occupy the leading place. Common here is the melt hardening at the surface of a rotating drum-cooler. In spinning, the molten metal is produced by induction or plasma-arc heating.

A laboratory plasma-arc furnace has been manufactured at the E.O. Paton Electric Welding Institute of NASU, which is designed for producing rapid-hardened alloys using a spinning method. The plasma-arc heat source makes it possible to realize a stable process of melting and drain of the melt of refractory alloys and non-compact charge, providing a wide range of the power control.

However, the production of these unique alloys remains mainly at the level of the laboratory production [1]. The basic difficulties, encountered in service of equipment, are the high amount of heat which is necessary to remove from the melt for a short interval of time, that is rather complicated to fulfill even in the laboratory equipment taking into account the rapidness of the process. Therefore, to design the industrial high-efficiency equipment, it is necessary to carry out a number of theoretical and experimental investigations which will help to make the preliminary evaluation.

It is quite difficult to make the experimental confirmation of mathematical models of heat transfer in spinning. Measurement of temperature inside the drum-cooler, rotating at high speed, is the insoluble problem using devices of pyrometry, and the contact methods of measurement are also complicated. As a rule, the impossibility of measurements is due to the fact that the rotating cooler has a complex inner shape (bores or channels of cooling) and in a majority of cases it is filled with any cooling agent (water, liquefied gases, etc.) [2].

There is no information in literature, devoted to the technology of a super-rapid hardening, about any

temperature measurements, however, there is measurement of temperature of moving surfaces in a metallurgical practice, some procedures are even described.

As a result of analysis of reference data, the idea was taken on the basis of a measuring sensor, rigidly fastened on the surface of object being examined, information from which is transmitted by flexible connecting wires using contact rings and sliding contacts to stationary set measuring devices.

Contact rings are widely used first of all in schemes with resistance thermometers (thermoresistors). But in all these systems it is difficult to eliminate the presence of parasitic thermoelectromotive forces and also changing contact resistances. Due to relative movement and clamping force between a contact brush and contact ring a thermoelectromotive force is generated, caused by a friction, which depends to some extent on combination of used materials. Graphite-containing materials in a pair with copper or silver generate a high thermoelectromotive force, that is connected with a constant contact resistance and low friction. Similar material for manufacture of contact rings and brushes can be used only in case when the rates of a relative movement are low [3].

Thus, the circuits of supply current and measuring system should be made high-resistance and the changing contact resistance of contact rings will not almost influence the results of measurements. Thermoresistor was selected as a contact sensor for the temperature measurement.

To transmit the changing resistance from a contact sensor (thermoresistor) using contact rings, a transmitting mechanism, consisting of a textolite casing and a rotor, was manufactured (Figures 1 and 2). The casing was fixed to a framework of the installation and two current removing graphitized contact brushes with clamping springs were mounted on it. The rotor was centered and rigidly fixed by pins at the edge of a rotation shaft of the drum-cooler and two copper circular contacts were fastened on it. As isolators, three fluoroplastic washers were placed between cop-

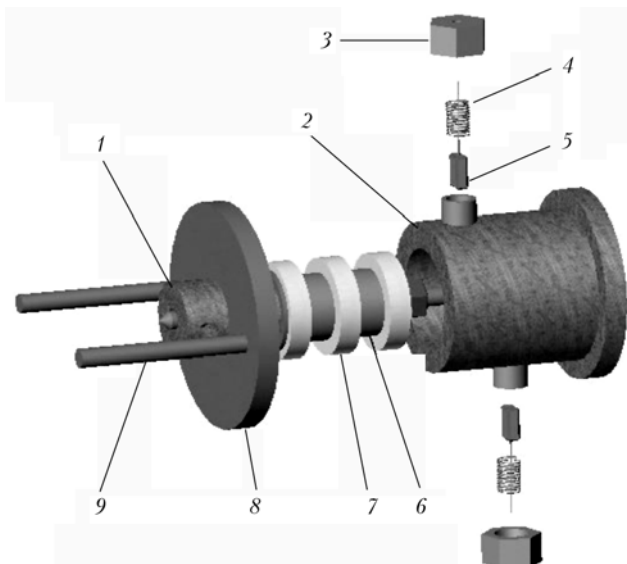


Figure 1. Mechanism for transmission of electric resistance of thermoresistor to measuring device: 1 — rotor; 2 — casing; 3 — fixing nut; 4 — clamping spring; 5 — graphitized brush; 6 — copper circular contacts; 7 — fluoroplastic rings; 8 — clamping washer; 9 — fastening pins

per contacts, the edge washers served also as sliding bearings and centered the casing with respect to the rotor rotation axis. Wires from copper circular contacts were wired inside the rotor and then connected to the thermoresistor. This scheme within the entire range of the drum-cooler rotation showed the contact resistance of not more than 100 Ohm.

Thus, to reduce the effect of contact resistances, it is necessary to select a sensitive element with a rated resistance by one order higher than the available one.

The present electronic industry produces a large assortment of thermoresistors, covering a large range of resistances suitable for measurement of temperature of up to 300 °C. Thermoresistors with a low inertia have a small size (bead-like). Their sensitive element is soldered into a glass drop. For these measurements a thermoresistor ST1-19 was selected (Figure 3), representing a non-insulated monolithic device of a direct preheating with a negative temperature coefficient of resistance. It is designed for measurement and control

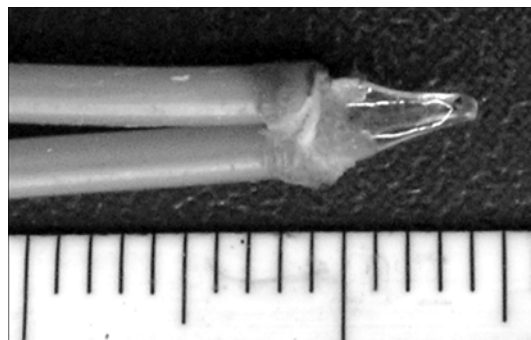


Figure 3. Sealed thermoresistor ST1-19

of temperature in electric circuits of direct and alternating currents of up to 400 Hz frequency.

To clarify the relation $R(T)$ of the selected thermoresistor, it was calibrated using laboratory thermometers TL-4, GOST 2823-59, with a scale graduation 0.1 °C (from 0 to 50) and 1 °C (from 0 to 100 °C). Dependence of resistance of thermoresistor on change in temperature is given in Figure 4.

Thermoresistor was fixed rigidly to the inner part of wall of the drum-cooler, whose internal cavity was filled with water (0.5 l) in the process of spinning for the more intensive cooling. During fastening of the sensor, all contact wires were insulated and connections, subjected to contact with water, were covered by a water-proof adhesive.

To record the indications of thermoresistors, the digital videofilming of ohmmeter and drain nozzle in one frame was made. The frequency was 30 frames per second.

Using the described procedure, measurements of temperature were made inside the rotating drum (Figure 5), where the beginning of coordinates coincided with a beginning of melt drain, point *a* was the drain completion. As is seen from the diagram, the system did not attain the temperature equilibrium due to a rapidness of the process and small volume of the drained metal.

To study the effect of heat loads on the drum-cooler during a long time, it is necessary to increase the volume of pouring metal, that is difficult to make in the laboratory conditions. Analogy was made with a continuous high-speed casting of cast iron between

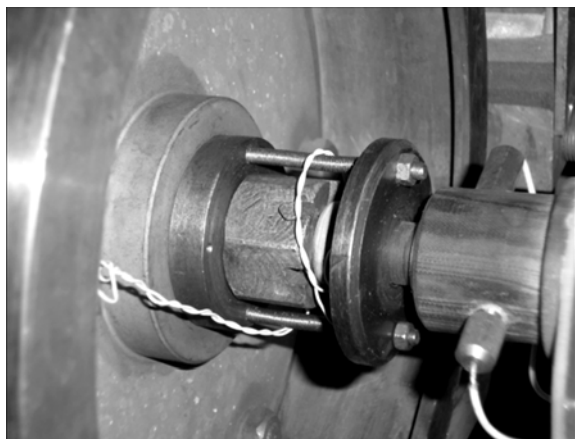


Figure 2. Fastening of a transmitter to the drum-cooler shaft

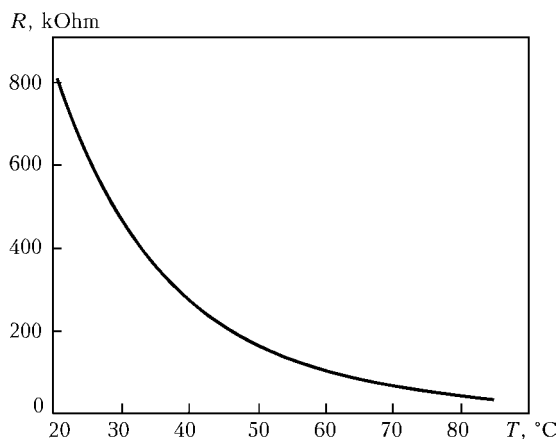


Figure 4. Temperature dependence of specific resistance R of thermoresistor ST1-19



two rolls into a sheet of 1.3–1.5 mm thickness. In this case the thermal equilibrium began after 15–20 s, and the temperature of rolls reached 350–370 °C.

As our preliminary calculations showed, the thermal equilibrium begins after 15–20 s, but at lower temperature (160–180 °C). This is connected both with lower temperature of melting of processed alloy of Ni–Si–B system (1020 °C), as compared with cast iron, and also with the fact that the drum-cooler is made of the more heat-conductive material (copper).

Thus, it is possible to make conclusions about heat inputs into the drum-cooler in hardening of a metal strip on it. The drum-cooler represents a hollow cylinder, inside of which a water is poured (as a cooling agent), which is distributed uniformly under the action of centrifugal forces during rotation. After determining the volume of a copper part of the cooler and water volume, it is possible to calculate the amount of absorbed heat. However, it is necessary to make some assumptions: temperature of copper and water is similar in all points (owing to high heat conductivity of materials); process is proceeding without heat losses into the environment; water volume is not changed.

The absorbed heat is equal to

$$Q = Q_c + Q_w = C_c M_c (T_f - T_b) + C_w M_w (T_f - T_b),$$

where Q_c is the absorbed heat by copper; Q_w is the absorbed heat by water; $C_c = 0.39$ kJ / (kg·deg) is the

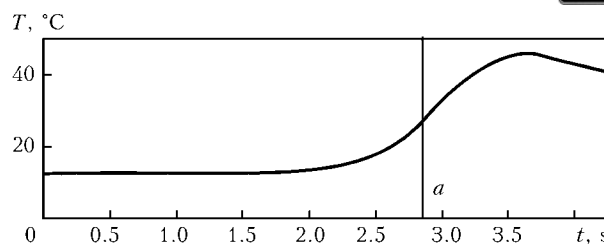


Figure 5. Change in temperature of an internal wall of drum-cooler in the process of spinning

heat capacity of copper; $C_w = 4.19$ kJ / (kg·deg) is the heat capacity of water; $M_c = 11.145$ kg is the copper mass; $M_w = 0.5$ kg is the water mass; $T_b = 12.6$ °C is the beginning temperature before drain; $T_f = 43.6$ °C is the maximum final temperature.

Thus, the heat energy absorbed by a copper drum-cooler is approximately 200 kJ.

As the experiments showed the carried out measurements give a relative idea about heat transfer in the spinning process and can be used in future for evaluation of accuracy of the developed mathematical models.

1. Zhadkevich, M.L., Shapovalov, V.A., Torkhov, G.F. et al. (2003) Producing of amorphous and nanocrystal materials using plasma-arc heating (Review). *Advances in Electrometallurgy*, 4, 28–33.
2. Abramenko, A.N., Kalinichenko, A.S., Bergman, G.V. et al. *Device for producing of metallic bands*. USSR author's cert. 1526892. Int. Cl. B 22 D. Publ. 07.01.93.
3. Lineveg, F. (1980) *Measurement of temperatures in engineering*.

TECHNOLOGIES FOR MELTING TITANIUM INGOTS FROM UNDERSEPARATED TITANIUM SPONGE IN SECTIONAL MOULD USING HIGH-FREQUENCY HEATING

Cutting down of costs in production of titanium sponge and its remelting into ingots is a major technical challenge. One of the methods to handle it is to shorten by 20–30 % the time of the processes of magnesium-thermic reduction of titanium from its chlorides with subsequent topping of the reaction products in vacuum (separation). This allows a 20 % decrease in consumption of power supplies and extension of service life of expensive equipment used for production of titanium sponge. The sponge produced by the shortened cycle contains up to 1 wt.% of chlorine compounds, instead of 0.08 wt.% required by the standard. Titanium sponge with an increased content of chlorides can be remelted into ingots using the process of induction melting in a sectional mould. The method is based on the fact that the zone with a high-intensity magnetic field is formed in the cooled copper mould using a small-height inductor, and that melting and refining of metal fed in portions during melting occur in this zone. To provide the incremental formation of an ingot, the zone is moved along the mould or left fixed, and then the ingot is drawn from the mould.

Purpose and application. Melting of titanium ingots from gas-saturated titanium sponge, lumpy scrap, wastes of pipe and plate rolling, chips.

State and level of development. Pilot plant is available for melting ingots of commercial titanium from underseparated titanium sponge. The metal meets specification requirements for chemical composition.

Proposals for co-operation. Building of the melting facilities and transfer of the technology to enterprises producing titanium.

Principal performers and developers: Shapovalov V.A., Dr. of Techn. Sci.; Shejko I.V., Dr. of Techn. Sci.; Konstantinov V.S., Cand. of Techn. Sci.

Contacts: Prof. Shapovalov V.A.
E-mail: shapovalov@paton.kiev.ua



DIFFUSION COATINGS ON THE BASE OF TITANIUM, VANADIUM AND CHROMIUM CARBIDES ON STEEL U8A

V.G. KHIZHNYAK¹, Yu.M. POMARIN², N.A. KURILO¹ and I.Yu. MEDOVA¹

¹NTUU «Kiev Polytechnic Institute», Kiev, Ukraine

²E.O. Paton Electric Welding Institute, NASU, Kiev, Ukraine

Phase composition, structure, microhardness, crack resistance of carbide coatings of titanium, vanadium, chromium on steel U8A are investigated. Experimental data about wear resistance of steel U8A with carbide coatings, obtained during tests under different conditions of friction are given: sliding friction without lubricant with and without preliminary grinding of surface and friction by a fixed abrasive. It was established that the carbide coatings increase the wear resistance of steel U8A by 2.5–20 times.

Key words: coatings, carbides, titanium, vanadium, chromium, steel, microhardness, crack- and wear-resistance

Modern machine parts, tools should have a complex of the following properties, providing their reliable service: high hardness, chemical stability in the wide range of temperatures, sufficient wear resistance. Realization of these properties can be solved by using coating–base metal compositions [1, 2]. Carbides, nitrides, borides of transition metals of IV–VI groups of the Mendeleev Periodic System of Elements can be referred to the most challenging materials of protective coatings. Except the above-mentioned properties, these materials should guarantee a stable operation of coating–steel U8A composition in the process of service. The important problem is the optimizing of not only structure and properties of the coatings, but also the conditions of their service [3–5]. For proper description of processes of deformation and fracture, it is necessary to use, except such characteristics of materials as modulus of elasticity, microhardness, the characteristics of crack resistance as well [6–9]. It should be noted that the processes of a brittle fracture of coatings on the base of interstitial phases, to which carbides of transition metals also belong, are defined by many factors. In spite of the achievements of mechanics of a contact fracture, there are no until now methods for determination of crack resistance of thin layers and procedures for differential evaluation of definite properties effect on performance of coating–base metal compositions. Thus, the determination of interrelation between structure, microhardness, crack resistance of the coating, its strength of adhesion with a base and wear resistance causes an indubitable interest. The obtained results allow proper selection of rational types of coatings for definite conditions of service.

Carbide coatings with content of titanium, vanadium, chromium were deposited on surface of steel U8A at 1323 K temperature and 8 h holding in a closed reaction space at low pressure [10]. Coatings were formed as a result of proceeding in a reaction

space of chemical transport reaction with participation of chlorides of transition metals. The latter are transporting the saturating materials to the steel surface, which interact with carbon of a gas phase and carbon of steel with the formation of carbides. Powders of transition metals, carbon tetrachloride and carbon-containing additions, were used as initial components. Phase and chemical composition of coatings were determined in X-ray diffractometer DRON 2.0 in a copper monochromatic radiation and analyzer «Camebax SX50». Metallographic examinations were carried out in microscope «Neophot 21».

Microhardness was determined in unit PMT-3, equipped with a device for record of a load–deformation curve. Characteristics of micro-strength, crack resistance of coatings were determined using procedure, described in work [11], at analysis of a curve of indentation by a tetrahedral diamond pyramid of unit PMT-3. In case of the coating fracture and formation of a crack of length C , an area is formed on load–deformation curve at load P_c , that allows determination of micro-strength by the following expression:

$$\sigma_{ms} = \frac{P_c}{C^2}.$$

The use of σ_{ms} as characteristic of strength is quite lawful. The crack is initiated in that case when the wedge stress exceeds the ultimate strength and is growing until the stress in the crack mouth becomes lower than the ultimate strength. By conventional features the carbide coating–steel U8A is characterized at the boundary by adhesion of the coating with a base.

To evaluate the crack resistance, a dimensionless value, named as a characteristic of microbrittleness γ , was used. At known load of crack initiation P_c , crack length C , load on indenter P_1 and diagonal of indentation d_1 the characteristic of microbrittleness was determined by the following expression:



Phase composition and properties of carbide coatings on steel U8A

Type of treatment	Phase composition	Thickness, μm	Microhardness, GPa	Load of crack formation P_c , N	Sizes of crack C , μm	Micro-strength σ_{ms} , MPa	Characteristic of micro-brittleness γ	Stress of coating separation, σ_{sep} , MPa
Titanizing	TiC	16.5	39.5	0.50	39	328.7	120.4	61
Titanium-vanadium-chromizing	(Ti, V, Cr)C	18.0	34.5	0.55	35	449.0	76.9	120
Vanadizing	V_2C	4.0	21.0	—	—	—	—	115
	VC	12.0	28.2	0.40	32	390.6	65.6	
Vanadium-chromizing	(V, Cr)C	13.5	23.7	0.50	30	555.6	42.7	180
	(Cr, Fe, V) $_7C_3$	3.0	17.5	—	—	—	—	
Chromizing	$Cr_{23}C_6$	7.0	16.0	0.75	33	687.7	23.2	220
	Cr_7C_3	11.5	17.1	0.70	34	605.5	28.5	

$$\gamma = 1.854 \frac{P_1 C^2}{P_c d_1^2}.$$

Thus, the characteristic of microbrittleness can be considered as effective characteristic of coating material, reflecting the ratio between the material microhardness (P_1 / d_1^2) and microstrength (P_c / C^2) for a definite stressed state.

The results of X-ray diffraction and metallographic examinations showed the following. The diffusion coatings, produced in this work, can be divided in two groups by a number of carbide layers. In titanizing, titanium-vanadium-chromizing the single-layer coatings on TiC base are formed on steel U8A. Content of vanadium and chromium in layer (Ti, V, Cr)C does not exceed, respectively, 0.8 and 2.7 wt.%. In vanadizing, vanadium-chromizing or chromizing the two-layer coatings of type VC, V_2C ; Cr_7C_3 , (V, Cr)C; Cr_7C_3 , $Cr_{23}C_6$ are formed. The mass fraction of chromium in layer (V, Cr)C for the accepted conditions of saturation is 10.0–13.5 %.

Carbide layers on steel U8A are determined after etching by 3 % solution of nitric acid in ethyl alcohol in the form of a light band with a distinct carbide–steel U8A boundary. Boundary between separate carbide layers is observed either by different methods of polishing or after etching by the reagent of Murakami.

Data about thickness, microhardness, characteristics of crack resistance of coatings are given in the Table.

Using the metallographic analysis of the coating material in the area of penetration of the diamond indenter of unit MPT-3, the presence of a zone of plasticity and zone of fracture, consisting of lateral and radial cracks was observed. Analysis of obtained results showed that coatings with chromium content have the highest adhesion with a base among the studied types of coatings. The above-mentioned coatings are typical of a good developed transition zone in the material base with an increased content of chromium. Besides, the base iron in a large amount is present in the zone of carbides. Thus, the mass fraction of iron

in the center of layer Cr_7C_3 is 47.6 and 42.8 %, respectively, in chromizing and vanadium-chromizing.

Results of investigations of load of a crack initiation, characterizing the capability of material of coating to elastic-plastic deformation before the moment of fracture, showed its adequacy with a microhardness. Thus, the lowest values P_c are established for coatings VC, whose microhardness is lower than that of coatings TiC, (Ti, V, Cr)C.

Mean values of load in case of formation of cracks and large sizes of cracks, which lie within the rather narrow interval for the investigated coatings, defined the low microstrength and high characteristic of microbrittleness of coatings TiC and (Ti, V, Cr)C.

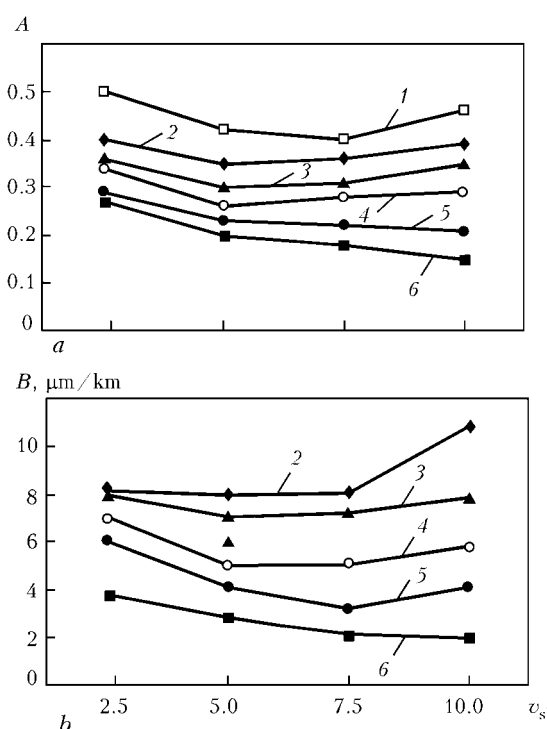


Figure 1. Characteristics of wear resistance of steel U8A with carbide coatings: *a* — dependence of friction coefficient A on sliding rate v_{sl} ; *b* — dependence of given wear B on v_{sl} : 1 — U8A; 2 — Cr_7C_3 , $Cr_{23}C_6$; 3 — VC, V_2C ; 4 — TiC; 5 — (V, Cr)C; 6 — (Ti, V, Cr)C; $P = 0.5$ MPa; given wear of U8A at sliding rates of 5 and 10 m/s is, respectively, 69 and 93 $\mu\text{m/km}$

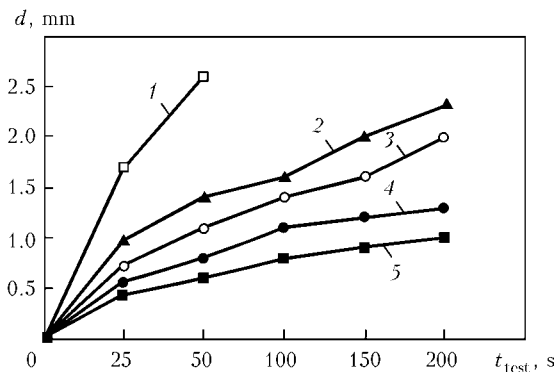


Figure 2. Wear resistance of steel U8A with carbide coatings at sliding friction without lubricant: 1 — U8A; 2 — Cr_7C_3 , Cr_{23}C_6 ; 3 — TiC; 4 — (Ti, V, Cr)C; 5 — VC, V_2C ; rate of sliding is 4 m/s, contact load of 1 N; d — size of dent

Results of investigations of wear resistance of steel U8A with carbide coatings at different conditions of friction surface contact are given in Figures 1–3.

Results of investigations by the shaft–insert scheme under sliding friction conditions without lubricant with a preliminary grinding of surfaces allowed classify the investigated coatings by increase in wear resistance and decrease in coefficient of friction as follows: Cr_7C_3 , Cr_{23}C_6 –VC, V_2C –TiC–(V, Cr)C, Cr_7C_3 –(Ti, V, Cr)C. High level of wear resistance of coatings (Ti, V, Cr)C; (V, Cr)C, Cr_7C_3 is defined by a high microhardness, by value of which they are inferior to titanium carbide TiC, by a mean crack resistance, a good adhesion with a base, as well as peculiarities of structure. These coatings consist of equiaxial grains, whose size for carbides (Ti, V, Cr)C and (V, Cr)C is, respectively, 0.1–0.2 and 0.3–0.4 μm , they are almost not textured. It can be noted that the microhardness does not define the high wear resistance of these coatings. In the zone of contact the coating at selected definite test conditions can reach the 800–1000 °C temperature [12, 13]. With increase in temperature the microhardness of carbides is decreased and already at 400–700 °C the hardness of carbides of vanadium, chromium occurs to be higher than the hardness of the titanium carbide TiC [14, 15]. Secondary structures, formed in the zone of contact of friction surfaces, influence greatly the wear resistance [16]. A layer of oxide, formed on the surface of titanium carbide TiC during friction, isolates effectively the friction pairs [1, 14].

Tribotechnical tests under conditions of a sliding friction without lubricant (Figure 2) by the shaft–insert scheme without a preliminary grinding of surfaces showed advantage of coatings on base of vanadium VC, V_2C and complex coatings (Ti, V, Cr)C. High rate of wear at the beginning of the experiment is common for all coatings during tests by the proposed procedure. This is caused by high contact loads at the beginning of the tests, which are decreased with increase in a dent, and, thus, promote decrease in the wear rate.

In addition, during tests by a proposed scheme the compressive stresses and tensile stresses in the direct vicinity, which can lead to separation of the

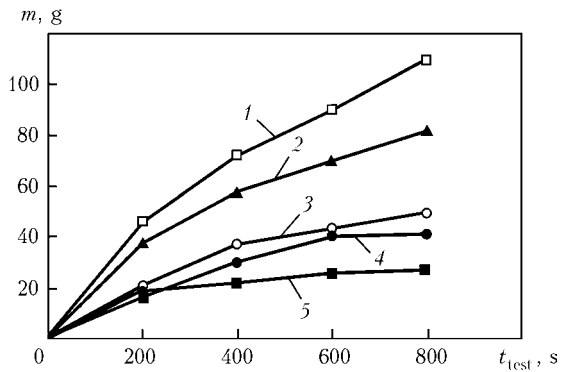


Figure 3. Abrasive resistance of steel U8A with carbide coatings: 1 — U8A; 2 — VC, V_2C ; 3 — TiC; 4 — (Ti, V, Cr)C; 5 — Cr_7C_3 , Cr_{23}C_6 ; m — mass loss

coating area, are acting in the zone of contact normal to the surface [6, 17]. Similar phenomena take place under the action of compressive stresses, occurring in coatings at cooling from temperatures of thermochemical treatment [2, 10, 18].

Thus, it can be considered that wear resistance of coating–steel combination under the given test conditions is defined by high microhardness and sufficient adhesion strength of coating with steel.

Tests by a fixed abrasive (Figure 3) showed that the carbide coatings 1.5–2.5 times increase the wear resistance of steel U8A. It was established that a coating in the zone of contact with abrasive is fractured in the first 180–240 s of test. Further fracturing of coating–steel combination is occurred at a constant rate. In accordance with the present representations, the abrasive wear resistance depends both on hardness of abrasive and base, and also on structure [19]. Moreover, homogeneous base has advantage over heterogeneous one under condition that ratio of abrasive hardness to base hardness does not exceed 1.2–1.3 [19]. Ratio of abrasive hardness (it is boron carbide in this work) to coating hardness is within 1.28–3.10. It can be considered that at abrasive wear the carbide layer–steel U8A combination is under such conditions at which some other characteristics, except hardness, should be taken into account.

CONCLUSIONS

1. Using the methods of a kinetic microhardness, the characteristics of microstrength, adhesion strength of coatings on the base of carbides of titanium, vanadium and chromium with steel U8A were determined. With increase in value of microstrength, the coatings are arranged in the row TiC, VC, (Ti, V, Cr)C, (V, Cr)C, Cr_7C_3 , Cr_{23}C_6 , as to crack resistance — TiC, (Ti, V, Cr)C, VC, (V, Cr)C, Cr_{23}C_6 , Cr_7C_3 , as to adhesion to base — TiC, VC, (Ti, V, Cr)C, (V, Cr)C, Cr_7C_3 , Cr_{23}C_6 .

2. It was established that the carbide coatings 2.5–20 times increase the wear resistance of steel U8A.

1. Samsonov, G.V., Epik, A.P. (1973) *Refractory coatings*. Moscow: Metallurgiya.
2. (1981) *Chemical heat treatment of metals and alloys*: Refer. Book. Moscow: Metallurgiya.



3. Khizhnyak, V.G., Korol, V.I. (2003) Stability of carbide-chrome coatings on steels at higher temperatures. *Meta-loznavstvo ta Obrob. Metaliv*, **3**, 24–29.
4. Bose, K., Wood, R. (2005) Optimum tests conditions for attaining uniform nolling abrasion in ball crotening tests on hard coatings. *Wear*, 322–333.
5. Gille, G. (1986) Strength of thin films and coatings. *Festkoperphys und Werkstofforsch*, **32**, 267–325.
6. Kolesnikov, Yu.V., Morozov, E.M. (1989) *Mechanics of contact fracture*. Moscow: Nauka.
7. Enmoto, Y. (1981) Sliding fracture of soda-line glass in liquid environments. *J. Mater. Sci.*, **16**(12), 3365–3370.
8. Sorokin, G.M. (1984) On nature of wear resistance of steels in abrasive wear. *Vestnik Mashinostroeniya*, **12**, 25–27.
9. Khizhnyak, V.G., Korol, V.I. (2002) Mechanical properties of carbide coatings with participation of chrome and steel U8A. *Fizyka i Khimiya Tv. Tila*, **1**(4), 161–165.
10. Loskutov, V.F., Khizhnyak, V.G., Kunitsky, Yu.A. et al. (1991) *Diffusion carbide coatings*. Kiev: Tekhnika.
11. Khizhnyak, V.G., Dudka, A.I., Khizhnyak, O.V. (1996) Determination of microbrittleness of carbide coatings using the methods of kinetic microhardness. *Izvestiya Vuzov. Chyorn. Metallurgiya*, **9**, 83.
12. Mamykin, E.T., Kovpak, M.K., Yuga, A.I. et al. (1973) Complex of machines and methods of determination of anti-friction properties of materials in sliding friction. *Poroshk. Metallurgiya*, **1**, 67–72.
13. Bernshtejn, M.L., Zajmovsky, V.A. (1977) *Mechanical properties of metals*. Moscow: Metallurgiya.
14. Tot, L. (1974) *Carbides and nitrides of transition metals*. Moscow: Mir.
15. Andrievsky, R.A., Spivak, I.I. (1989) *Strength of refractory compounds and materials on their base: Refer. Book*. Chelyabinsk: Metallurgiya.
16. Kostetsky, B.I. (1989) *Wear resistance of machine parts*. Moscow, Kiev: Mashgiz.
17. Khizhnyak, V.G., Kostenko, A.D., Korol, V.I. (2003) Wear resistance of carbide and boride coatings on steel U8A. *Poroshk. Metallurgiya*, **11/12**, 110–115.
18. Powell, C.F. (1996) *Vapor deposition*. New York, 198.
19. Torrance, A.A. (2005) Modeling abrasive wear. *Wear*, **258**, 281–293.

STRUCTURE AND STRENGTH PROPERTIES OF BRAZED JOINTS OF CAST NICKEL ALLOY JS26NK

Part 1

I.S. MALASHENKO¹, V.V. KURENKOVA¹, A.F. BELYAVIN¹, V.V. TROKHIMCHENKO²,
L.V. CHERVYAKOVA² and Ya.B. LEBEDEVICH²

¹Research Center «Pratt & Whitney Paton», Kiev, Ukraine

²E.O. Paton Electric Welding Institute, NASU, Kiev, Ukraine

Materials science specifics of formation of brazed joints of cast nickel alloy JS26NK is considered. Physical-mechanical properties of the brazed joints produced by using standard boron-containing brazing alloy and brazing alloy, having silicon as the second depressant, as well as different binders, are given. It was shown that the use of a binder in the form of solution of acrylic acid in acetone is effective in case of application of composite brazing alloys containing 20 % of powder of the industrial alloy Ni–12 % Si (NS12). Strength and fracture toughness of the brazed joint of alloy JS26NK, attained at room temperature, exceed the similar characteristics of alloy JS26VI. Relation between mechanical properties of the brazed joints and direction of prevailed growth of dendrites in the brazed metal was demonstrated.

Key words: brazing in vacuum, cast alloy JS26NK, composite boron- and silicon-containing brazing alloy, brazing alloy NS12, brazed joint, strength, elongation, fracture, structure, fusion line

The directional crystallization (DC) of heat-resistant alloys found a wide spreading in manufacture of gas turbine blades. According to data of works [1, 2], the process of DC has been developed for the purpose that grains, characterized by a low value of modulus of elasticity and oriented along the crystallographic direction <001>, could grow along the axis Z of a blade. The absence in the blade of transverse grain boundaries in combination with a low value of modulus of elasticity guarantees the metal the 3–5-fold increase in life under conditions of a thermal fatigue, as compared with castings, having an equiaxial structure.

In work [3] the physical-mechanical properties of brazed joints (BJ) of alloy JS26VI at room and high temperature are described. Samples for tests were made from brazed plates of an equiaxial crystallization using 40 wt.% of standard brazing alloy #1 (Ni–Co–

Cr–Al–2.5 % B system) + 60 wt.% of filler in the form of a powder of alloy Rene-142 and using a complex brazing mixture, in which 20 % of a standard brazing alloy #1 was replaced by 20 % of industrial alloy Ni–12 % Si (NS-12). To prepare the mixture, a special material «Binder», made in the USA, was used as a binder in this series of experiments.

In the present work the plates-billets were used, melted out at enterprise «Motor Sich» using DC. Chemical etching of plates showed that their macrostructure is greatly differed by a direction of growth of primary dendrites (Figure 1).

According to data of work [4], the anisotropy of strength properties of single crystals of nickel heat-resistant alloys (HRA) and different nature of diagrams of deforming are defined by the interaction of definite systems of slipping at loading (tensile test).

At room temperature the highest values of yield strength and ultimate tensile strength on the example of deformation of single crystals of alloy JS6F are typical of orientation <111>, demonstrating a negligible area of yielding at tension. Orientation <112>

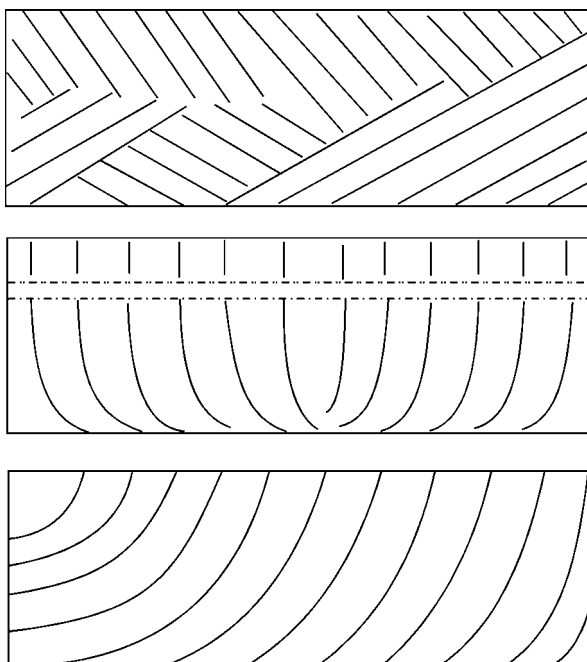


Figure 1. Direction of growth of dendrites in plates of JS26 alloy

is less strong. Ductility (elongation) of these single crystals does not exceed 10 %.

Single crystals with orientations, close to $\langle 100 \rangle$ and $\langle 111 \rangle$, are deformed by a multiple slipping along the crossing planes that causes their deformation strengthening (Figure 2). Tensile curves of samples with orientation $\langle 011 \rangle$ have an expressed peak of yielding with an extended stage of a slight slipping. In this case there are no almost the deformation strengthening. Slipping in a single crystal is realized by a single system of non-crossing planes. Maximum elongation of a single-crystal alloy JS6F due to a complete realization of this stage of slipping reaches 30 %.

Classic curve of a monotonous deformation strengthening (when yield strength corresponds approximately to 700 MPa, and ultimate tensile strength reaches 900 MPa at elongation of about 20 %) distinguishes single crystals of orientation $\langle 001 \rangle$ in tension from other orientations depending on conditions of solidification (heat removal). Typical examples of peculiarities of crystallization are shown schematically in Figure 1.

Data, given in Figure 2, allows better understanding of the nature of deformation strengthening of BJ of alloy JS26NK with a coarse-crystalline di-

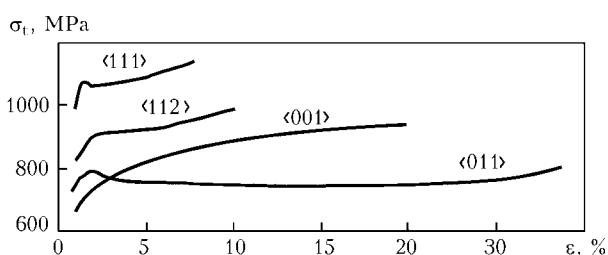


Figure 2. Typical diagrams of deformation of single crystals of alloy JS6F of different crystallographic orientation at 20 °C temperature [4]

rected structure, that is a subject of the present investigation.

Spectrum of diagrams of deforming single crystals of alloy JS6F, given in work [4], prove convincingly that some produced samples of BJ had orientation of growth, close to $\langle 001 \rangle$ or, possibly, $\langle 112 \rangle$, while ductility of BJ was negligible at an increased level of strength. During tension of single crystals with orientations $\langle 001 \rangle$ and $\langle 111 \rangle$ the crystalline lattice in samples is not almost turned and, due to this, the initial orientations are not changed.

There is a definite relationship between the ultimate tensile strength σ_t and orientation of crystals in alloy JS26NK [5]. Maximum value of ultimate tensile strength is characteristic of samples, in which the axis is close to direction $\langle 111 \rangle$. Favorable orientation of the growth of grains in castings is such an orientation at which the vector of applied load coincides with crystallographic direction $\langle 100 \rangle$ or $\langle 111 \rangle$.

Samples with a distinctly expressed orientation of a directed growth are characterized by different diagrams of deformation strengthening, the main characteristics of which are yield strength, nature of deformation strengthening, elongation (total ductility before fracture).

According to work [4], the anisotropy of mechanical properties, connected with a different dendritic structure of single crystals equivalent to crystallographic orientation, is called growing, as it is defined by deviation of a sample axis relative to the direction of the crystal growth.

Crystallographic and «growing» anisotropies depend on conditions of producing and testing of metal of castings, i.e. from conditions of crystallization, chemical composition of the casting and heat treatment. Size of particles of the strengthening γ' -phase, defined by the heat treatment condition, influences greatly the strength characteristics of metal of DC: with increase in mean size of particles of γ' -phase the values $\sigma_{0.2}$ and σ_t are reduced, and the ductility is growing.

The largest deviations in distribution of structure components in alloys of DC cause eutectic precipitations of carbides MeC, γ' -phases and carboroborides, as they are oriented randomly with respect to solidified matrix and have an arbitrary morphology. Distribution of particles of γ' -phase in matrix of DC is not also uniform by sizes. Their non-uniform distribution in matrix is connected with a dendritic liquation. High-temperature annealing at 1220 °C in vacuum levels partially the chemical composition of axes of dendrites and interaxial areas, promotes the precipitation of particles of γ' -phase, uniform by sizes and morphology, in decay of γ -solid solution in the process of cooling from the temperature of homogenization.

Materials and procedure of investigations. It was supposed in the course of testing that the strength of BJ can (and should) be different depending on the method of cutting out of plates with respect to ad-



vantageous orientation of growing dendrites in the brazed metal.

Material for investigations was initial cast billets of alloy JS26NK of 6.5 mm thickness. Each of them was polished and cut into two thin plates of approximately 3 mm thickness, and then they were annealed in vacuum $5.5 \cdot 10^{-3}$ Pa at temperature 1225 °C for 1 h. Vacuum etching of a polished surface confirmed that the metal structure was formed under the conditions of oriented solidification, i.e. plates became of DC.

At the first stage of experiments two halves of the plate were split into pieces for manufacture of samples [6] without a precise dependence on growth direction (Figure 1, c), then they were subjected to «wet» polishing for removing an oxide film after their annealing in vacuum.

Brazing alloy mixture was selected coming from results of previous investigations, carried out on alloy JS26 of an equiaxial crystallization [3]. Two types of mixtures were used: #1 + 60 % Rene-142, and #1 + 20 % NS12 + 60 % Rene-142. Best mechanical properties were obtained on samples of BJ of alloy JS26VI, formed with use of the above-mentioned system of composite brazing alloys.

Type of a binder served a variable factor in the conductance of an initial experiment: both a standard organic (Binder-215) and also a solution of acrylic resin in acetone (AA) were used. Both types of binders were tested on two types of composite brazing alloys.

On conjugated areas of plates being brazed a powdered brazing mixture was deposited in the form of a bead, this mixture was impregnated with a solution of an acrylic resin or mixed using Binder-215. The assembled halves were pressed at a low force and dried. At such a scheme of preparation of BJ, the natural gap between the contacting plates was minimum, i.e. approximately 125–200 μm .

At the second stage of investigations the preparation of billets of alloy JS26NK for brazing was made depending on the direction of growth of crystals. In the initial billet the direction of growth of dendrite in four samples coincided with a long axis of the plate. Six «transverse» pieces of BJ were cut out from the second plate. In this case the direction of growth of dendrites was coincided with the brazed weld axis (Figure 3). It was supposed initially that such geometry does not guarantee the high strength of BJ, as in brazing the diffusion processes could decelerate at the boundary of «metal being brazed–brazing alloy melt» contact because of the absence of boundaries of adjacent dendrites, which could facilitate the penetration of components of the molten brazing alloy into a base metal. In practice, this hypothesis occurred to be erroneous, as the boundaries of branches of dendrites (of a higher order) were «working» at BJ loading.

In the above-mentioned experiments, the solution AA was used as a binder, while the composition #1 + 20 % NS12 + 60 % Rene-142 was used as a basic brazing alloy. Natural gap varied and was actually

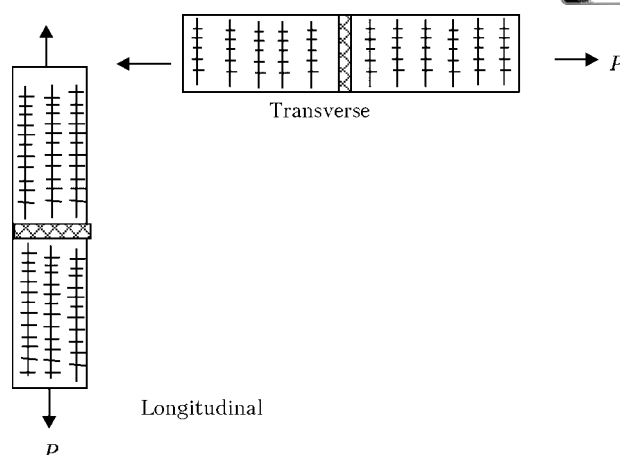


Figure 3. Plates of BJ of JS26NK alloy

350–420 μm for longitudinal samples and 500–780 μm for transverse samples.

The isothermal brazing in vacuum was performed at temperature 1225 °C (15 min), then the samples were annealed in vacuum at 1160 °C for 2 h, and then subjected to ageing at 1050 °C (4 or 5 h).

Experimental results. At the first stage the strength of initial alloy JS26NK depending on orientation of growth of dendrites was not investigated because of a deficit of the material. The appropriate data are given in Table 1 for the BJ. The performed brazing provided a quality interaction of weld metal and metal being brazed, owing to which the fracture of BJ was mainly in base metal (along the fusion line). Therefore, in separate cases the test result was taken as a strength of the metal being brazed, having a definite crystallography. Here, the strength of weld metal exceeded the value of ultimate tensile strength of the brazed alloy (Table 1).

With account for the chemistry of the brazing alloy, used in the technology of brazing, it can be stated that the application of solution AA as a binder is more effective in work with brazing alloy containing silicon. This brazing alloy allowed producing BJ, whose ultimate tensile strength exceeded 1000 MPa at stable high value of elongation of 9–17 %. At the same time, the solution of acrylic resin occurred to be so effective binder, when a basic brazing alloy without NS12 was mixed. Brazing alloy, containing silicon and boron simultaneously as depressants, provided higher technological effectiveness of BJ of alloy JS26NK.

Elongation of sample Z8, obtained using a basic brazing alloy, occurred to be maximum (23 %). This sample became exception from the general results of tests.

The first stage of experiments demonstrated stable good results on a short-time strength and ductility of BJ, made using a composite brazing alloy #1 + 20 % NS12 + 60 % Rene-142 (Table 1). Soundness of BJ in this case was 80–130 %.

Depending on the crystallography of growth of dendrites in plates being brazed, curves of deformation of samples different by nature, were obtained (Fi-

**Table 1.** Mechanical properties of BJ of JS26NK alloy at room temperature (brazing Nos. 38 and 40)

Conditions of brazing and heat treatment	No. of sample	Brazing alloy	$\sigma_{0.2}$, MPa	σ_t , MPa	ϵ , %	Gap, μm	Binder	Cut out of samples		
1225 °C, 15 min 1160 °Ñ, 2 h + 1050 °C, 4 h	Z0	40 % #1 + 60 % Rene-142	848.0	848.7	0	405	8 % AA	Arbitrary		
	Z2		664.5	723.6	6.25	—				
	Z4		683.7	697.7	0.7	335				
	Z6*		740.0	822.3	1.2	225	Binder			
	Z8		688.8	853.0	23.0	125				
	Z12		650.3	743.0	14.8	—				
	Z1	20 % #1 + 20 % NS12 + 60 % Rene-142	652.0	1067.0	12.5	270	8 % AA			
	Z3*		657.2	823.0	17.3	—				
	Z5*		753.2	1065.0	8.7	220			Binder	
	Z9		673.6	760.0	15.5	230				
	Z11		613.0	615.8	0.85	200				
	Z13		623.3	690.8	5.0	200				
	1225 °C, 15 min 1160 °Ñ, 2 h + 1050 °Ñ, 5 h **	Z20	20 % #1 + 20 % NS12 + 60 % Rene-142	827.0	882.0	4.2	350		8 % AA	Longitudi- nal
		Z21		770.8	855.0	4.5	400			
Z22		835.7		881.0	3.9	400				
Z23		802.6		884.0	5.5	420				
Z1' (#16)		20 % #1 + 20 % NS12 + 60 % Rene-142	752.0	850.2	4.5	570	8 % AA	Trans- verse		
Z2' (#19)			734.7	877.0	8.2	500				
Z3' (#15)			747.2	766.5	0.8	500				
Z4' (#17)			680.8	890.0	9.5	630				
Z5' (#14)			739.4	926.8	10.5	780				
Z0' (#18)			738.7	801.7	2.8	660				

* Smooth samples, test part was not subjected to grinding (to thinning).
** Heat treatment after brazing without removal from furnace.

*Smooth samples, test part was not subjected to grinding (to thinning).

**Heat treatment after brazing without removal from furnace.

Figure 4). Among BJ, made using a composite alloy, the samples should be distinguished, characterized by the most advantageous combination of physical and mechanical properties (Table 2). To estimate objectively the obtained results, a single (best by values and stability) system of samples of BJ, formed using

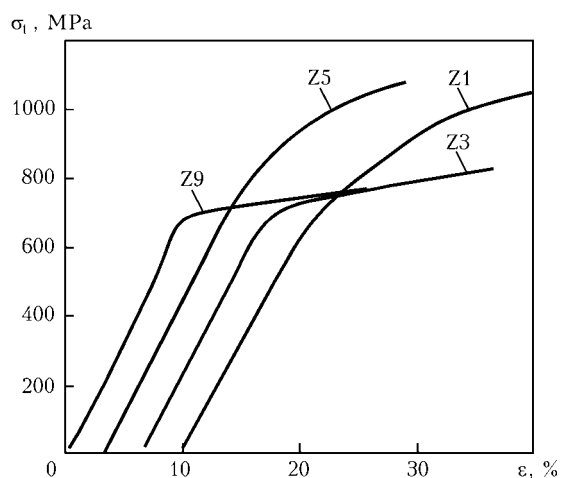


Figure 4. Diagrams of tension of samples of BJ of JS26NK / 20 % NS12 + 20 % #1 + 60 % Rene-142 / JS26NK of different crystallographic orientation fractured along the fusion line and base metal: Z9 and Z3 — orientation <001>; Z5 and Z1 — <111>

composite brazing alloy #1 + 20 % NS12 + 60 % Rene-142 and a binder on the base of acryl resin, was studied.

The more distinctly expressed absence of defects in weld metal, the higher homogeneity of its structure, the higher strength of BJ was reached.

Samples of BJ were characterized mainly by close values of yield strength within the interval of 623.3–688.8 MPa. The mean value $\sigma_{0.2}$ from the data of testing eight samples was 656.8 MPa, that is close to values of yield strength of alloy JS26NK. With account for values of ductility of BJ and nature of curves of deformation strengthening it is possible to state that the factor of crystallographic orientation of dendrites, composing the test part of samples, had a rather great influence on the level of physical-mechanical properties of BJ of alloy JS26NK.

Two of studied samples (Z1 and Z5) had a maximum value of ultimate tensile strength of 1065–1067 MPa, moreover, one of them had a standard yield strength $\sigma_{0.2} = 652$, another one — 753.2 MPa. Character of diagrams of tension of the mentioned samples proved intensive deformation strengthening of alloy with an approximate orientation of crystal-lites [111] (Figure 4).



Table 2. Generalized results of mechanical tests of BJ of JS26NK alloy produced from plates of different type after two-stage heat treatment

Yield strength, MPa	Ultimate tensile strength, MPa	Number of tests of samples, pcs	Structure of castings	Cut out of plates
611.2–622.3	694.3–742.2	10*	Equiaxial	Arbitrary
623.3–842.0	690.8–1067.0	12	Directed	Same
681.0–740.0	766.0–927.0	6	Same	Transverse
771.0–836.0	855.0–884.0	4	»	Longitudinal

*Two brazings, where a solution of acrylic resin was used as a binder.

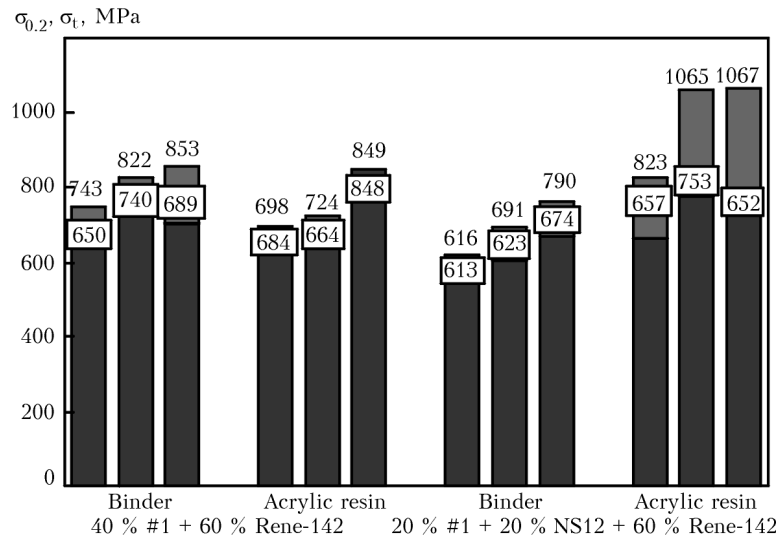


Figure 5. Strength of BJ of alloy JS26NK produced by the method of contact brazing in vacuum using boron-containing brazing alloys with and without addition of silicon in the form of powder of alloy Ni-12 % Si

Two other typical cases of deforming brazed samples Z3 and Z9 in tension are described by a curve with a steeper nature: at low values of strength (760–823 MPa) and low values of yield strength (657–674 MPa) the strengthening was almost absent, that corresponds to orientation close to [001] (Figure 4) [4].

Among twelve test samples, arbitrary cut out of a cast plate, nine samples had close values of yield strength of (667.2 ± 20) MPa (Tables 1 and 2, Figure 5), that corresponded to those for BJ with an equiaxial structure [3].

In study of a group of samples of BJ, formed using a composite brazing alloy #1 + 60 % Rene-142, the samples Z0 and Z8 with elongation equal, respectively, to 0 and 23 % are most polar (Figure 6). It is not correct to state about the possible effect of direction of growth of dendrites in plates on mechanical properties of these samples of BJ, as in this case the BJ had a different size of a gap (width of a brazed weld), such as 125 μm in sample Z8 and 405 μm in sample Z0. Maximum elongation (23 %) was obtained in sample Z8 at $\sigma_{0.2} = 689$ MPa, corresponding to yield strength of the alloy being brazed. At a small natural gap (125 μm) the volume fraction of brittle phases is minimum, owing to which the formed weld had a structure close to a matrix γ -solution, providing a high level of fracture toughness to it. In this sample the fracture occurred along the diffusion zone (Figure 7).

From group of BJ of Table 1, the sample Z0 was characterized by a high elasticity limit (848 MPa) at actually zero ductility. Sample had many brittle secondary phases along the boundaries of dendrites in a brazed weld, therefore it fractured by reaching the yield strength.

The main aim of the second stage of brazing the samples of alloy JS26NK was to determine the effect of method of cutting out of plates relative to direction of growth of dendrites in them on mechanical prop-

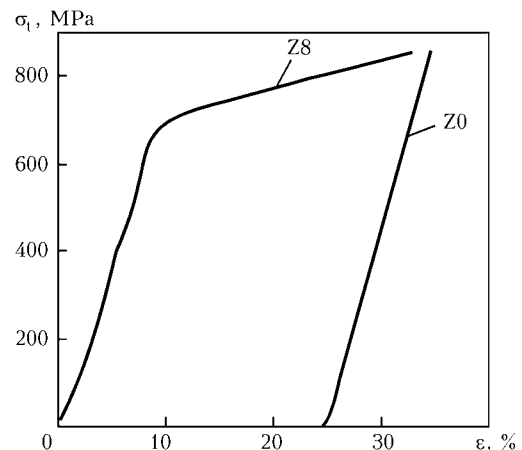


Figure 6. Diagrams of tension of samples of BJ of JS26NK / 40 % #1 + 60 % Rene-142 / JS26NK of different crystallographic orientation fractured along the fusion line and weld metal: Z8 — $\varepsilon = 23\%$; Z0 — $\varepsilon = 0$

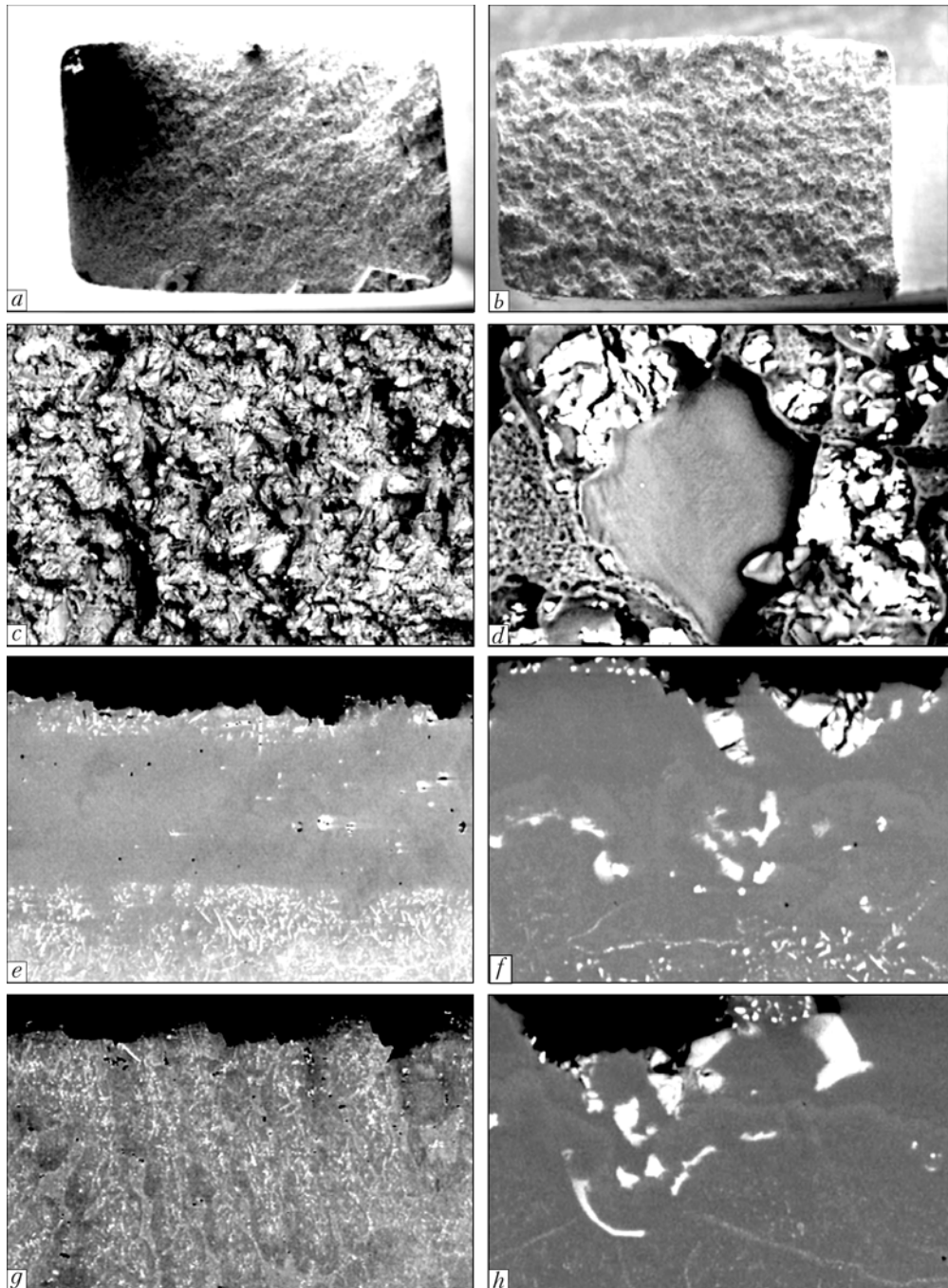


Figure 7. Pattern of tough (sample Z8, *a, c, e, g*) and brittle (sample Z6, *b, d, f, h*) fracture of BJ formed using a basic brazing alloy 40 % #1 + 60 % Rene-142 at a uniaxial tension: *a, b* — general view of fractures of samples, $\times 50$; *c, d* — patterns of tough and brittle fracture of samples, $\times 200$, $\times 1000$; *e, g* — two opposite fracture surfaces of sample Z8 ($\sigma_t = 853$ MPa, $\varepsilon = 23\%$), $\times 200$; *f, h* — fracture surface of sample Z6 in brittle phases in weld ($\sigma_t = 822$ MPa, $\varepsilon = 1.2\%$), $\times 1000$

erties of BJ (see Table 1). In transverse samples the boundaries of adjacent dendrites were arranged normal to the direction of the applied load. Fusion lines are parallel to axes of dendrites of the first order, however, owing to the presence of dendrites of the second and third orders and subsequent heat treatment the diffusion interaction between the metal being brazed and brazing alloy melt was realized to a full degree.

The obtained results allowed us to state that the essential difference in strength of samples of alloy JS26NK in two different directions of casting was not observed (Figure 8). It was supposed that the ductility of longitudinal samples in the direction of growth

of dendrites (normal to forming axis of weld) will be much higher.

Results of tests of BJ are given in Table 1 and Figure 8, from which some advantage of mechanical characteristics of longitudinal BJ as compared with transverse samples is seen.

In longitudinal samples Z20–Z23 the deformation strengthening is less expressed than in transverse samples, in which the steepness of a curve of loading is more visible as the total area of grain boundaries subjected to load, is larger.

The longitudinal samples had the higher yield strength ($\sigma_{0.2} = 802\text{--}835$ MPa) at ultimate tensile

**Table 3.** Strength of BJ of JS26NK alloy at high values of temperature in air

No. of sample	Brazing alloy	$\dot{O}_{\text{test}}, ^\circ\text{N}$	σ_t, MPa	$\epsilon, \%$	Remarks
ZK0	40 % #1 + 60 % Rene-142	950	470.0	0.25	—
ZK8		1000	407.5	0.35	
ZK5	20 % #1 + 20 % NS12 + 60 % Rene-142	950	477.0	0.30	—
ZK6		1000	427.4	0.45	
ZK7		1000	398.0	0.25	Additional ageing at 900 °C

Note. 8 % solution AA was used in all cases of a contact isothermal brazing of plates as a binder.

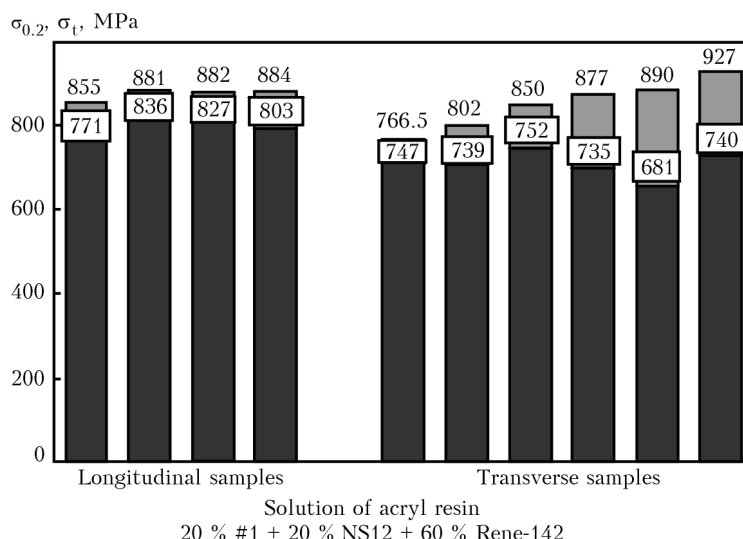


Figure 8. Strength of samples of BJ of JS26NK alloy produced by the method of a contact brazing in vacuum using a composite brazing alloy 20 % #1 + 20 % NS12 + 60 % Rene-142 in cutting out of pieces for brazing from plate along and across the direction of growth of crystallites

strength $\sigma_t = 881\text{--}884$ MPa. Evidently, this is connected with a long-time (5 h) heat treatment of BJ at 1050 °C.

Figure 9 shows statistic curves of processing of values of yield strength and ultimate tensile strength of BJ. Data of testing 22 samples were studied. After the two-stage heat treatment of BJ of alloy JS26NK in vacuum the yield strength of 73 % of joints was 675–775 MPa, and ultimate tensile strength of 67 % of samples was 775–925 MPa.

Analysis of results of testing the second series of experiments proved that mechanical properties of longitudinal and transverse samples of BJ have the close values.

A good repetition of values of yield strength of BJ sample, made from plates cut out across the direction of growth of dendrites, and ultimate tensile strength of longitudinal samples of BJ, was obtained (Figure 8).

Strength of transverse samples was 766–927 MPa, that is naturally associated with amount of dendrites, composing sample of BJ, and total area of the grain boundaries.

The peculiar feature of the second stage of experiments for brazing on alloy of DC was the use of homogenizing annealing (1160 °C, 2 h) of BJ without removal of samples and long-time (5 h) ageing at 1050 °C. This condition of heat treatment allowed

complete eliminating the clearly expressed boundaries (contour) of the weld fusion line. Transition between thermally-stable alloy and weld metal as a result of application of a composite alloy with silicon was less visible; in this case the mutual diffusion of components of alloy and brazing alloy occurred to be more complete.

At the second series of experiments the gap between the plate halves being brazed in BJ was much larger (350–780 μm) than in the first series (200–400 μm). It follows from this that at the larger width (2 times) of weld the elongation of joints became smaller (2.8–10.5 %). This proves that in case of a

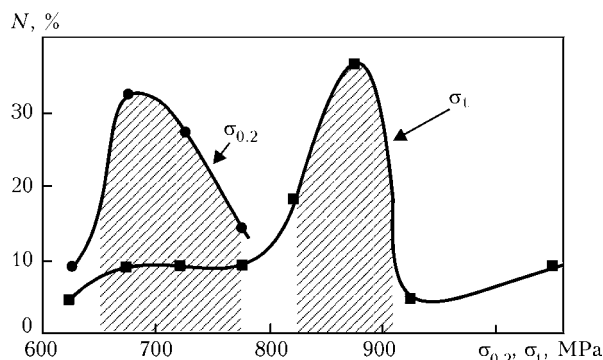


Figure 9. Statistic curves of distribution of values $\sigma_{0.2}$ and σ_t of BJ of JS26NK alloy produced using boron- and silicon-containing brazing alloys: N — number of samples



wide gap, i.e. large amount of used composite alloy, the determining ductility of samples characterizes the ductility of weld metal at the highest probability.

Fracture of samples occurred in microdefects of the solidified weld metal, as the precipitation of coarse or thin (tearing) carboboride phases was not observed in the diffusion zone and fusion line. At a small width of a gap (200–400 μm .) the fracture occurred along the fusion line or base metal. The composite alloy provided a high strength of welds of BJ, and the finishing heat treatment contributed to precipitation of strengthening phases of a different type along the fusion line. These factors influence the fact that samples with a small gap fractured in base metal, elongation of joints caused by deformation of base metal was high, i.e. 5.0–17.5 %.

In work [7] the negative effect of gap on mechanical properties of BJ was outlined. Thus, in brazing without a gap (weld of up to 100 μm) the amount of the secondary (brittle) phase in weld was minimum. With increase in gap width the amount and sizes of secondary components of the weld, became the potential sources of fracture of samples during loading, were growing.

The conductance of a long-time ageing at 1050 °C for wide-gap samples occurred to be justified as it allowed eliminating of variety and development of secondary phases in the solidified weld metal and providing leveling the metal composition at the base metal–brazing alloy interface.

In case of brazing of JS26NK alloy a high weakening at high temperature was observed in samples of BJ, formed using the brazing alloy not containing NS12 (Table 3). Refining (dispersion) of precipitations of the carboboride phase in adding of silicon into the basic brazing alloy made it possible to attain the increase in the level of strength of BJ at short-time tests in air at 950–1000 °C.

Final annealing at 1050 °C (4 h), being higher than test temperature, occurred to be preferring heat treatment (as compared with ageing at 950 °C, 3 h) and provided a real 1.5-fold increase in BJ strength.

1. Dal, D.N. (1995) *Superalloys of directional solidification*. Superalloys II: Refractory materials for aerospace and industrial power plants. Book 1. Moscow: Mir.
2. Ross, I.V., Sims, Ch.T. (1995) *Alloys on the nickel base*. Book 1.
3. Malashenko, I.S., Kurenkova, V.V., Onoprienko, E.V. et al. (2007) Mechanical properties and structure of brazed joints of casting nickel alloy JS26VI. Part 1. *Advances in Electrometallurgy*, **1**, 21–27.
4. Shalin, R.E., Svetlov, I.L., Kachanov, E.B. et al. (1997) *Single crystals of nickel refractory alloys*. Moscow: Mashinostroenie.
5. Yagodkin, Yu.D., Shulyak, V.P., Orekhov, V.B. (1987) Mechanical properties and orientation of crystals in samples of JS26 alloys melted by directional solidification. *Aviats. Promyshlennost*, **2**, 50–51.
6. Kurenkova, V.V., Malashenko, I.S., Trokhimchenko, V.V. et al. (2006) Short-term strength of brazed joints of nickel alloy VJL12U at temperatures 20 and 950 °C. *Advances in Electrometallurgy*, **3**, 27–37.
7. Kvasnitsky, V.F. (1986) *Welding and brazing of refractory alloys in shipbuilding*. Leningrad: Sudostroenie.

TECHNOLOGY AND EQUIPMENT FOR REFINING OF CRUDE RARE-EARTH METALS

Metal-thermic reduction is the most common method used to commercially produce the majority of REM. Calcium, aluminium or carbon are most often employed as reducing agents. Crude rare-earth metals produced by this technology contain from 0.7 to 2.0 % calcium. Therefore, they should be additionally refined in arc skull furnaces. This process is characterised by increased power and labour consumption.

The E.O. Paton Electric Welding Institute developed the technology and built the pilot unit for refining of crude REM, based on induction remelting with formation of ingots in sectional water-cooled mould. Intensive stirring of the metal melt in the high-frequency electromagnetic field provides a several times increase in the rate of evaporation of calcium from the metal melt, compared with traditional arc melting. The refining process is performed in the inert gas atmosphere, and pressure in the melting chamber can be varied over a wide range. The pilot unit allows melting of high-purity REM ingots with a diameter of 100 to 200 mm and weight of up to 80 kg. The annual output of the unit is up to 30 t.

Purpose and application. The technology is intended for refining of crude rare-earth metals having a melting point of not lower than 1200 °C. The technology and equipment have been applied at the Kirgiz Mining Works to produce pure yttrium of the Itm1 grade with a calcium content of up to 0.01 %.

State and level of development. Experimental-industrial verification at the Kirgiz Mining Works.

Principal performers and developers: Shapovalov V.A., Dr. of Techn. Sci.; Shejko I.V., Dr. of Techn. Sci.

Contacts: Prof. Shapovalov V.A.
E-mail: shapovalov@paton.kiev.ua



ANALYSIS OF THE STATE-OF-THE-ART AND PROBLEMS OF DEVELOPMENT OF CONTINUOUS CASTING OF STEEL IN METALLURGY OF UKRAINE

V.K. POSTIZHENKO and G.G. EFIMENKO
NTUU «Kiev Polytechnic Institute», Kiev, Ukraine

Review of development of structure of the process of a continuous casting of steel during recent 50 years is presented. The share of the continuous casting in countries with a developed steel making production in the 1990s approached 100 %. At the beginning of the XXI century a jump was made in the technology of steel production, thus marking the beginning of the industrial manufacture of hot-rolled strips using casting-rolling technologies which are priority now in steel casting.

Key words: casting of steel, processes of solidification of molten metal, machines for continuous casting, mould, quality metal products, physical and chemical heterogeneity of steel ingot, electromagnetic stirring

Steel casting, one of the main processing stages in metallurgy, is a process of a billet production for next rolling and producing of ready metal products (thin and thick sheets, rolled sections, rails, etc.).

The quality of the final metal products depends greatly on the initial billet. All processes of producing molten metal are directed to the improvement of its chemical composition, purification from non-metallic inclusions, gas components, etc.

However, to produce the quality metal products, it is necessary to take into account also the factor of metal crystallization.

Crystallization of the molten steel is governed by definite laws and is a rather complicated process. As regards to a steel ingot, it means, first of all, the suppression of physical and chemical heterogeneity, imparted to appropriate metal products.

The higher homogeneity of the metal products, the lower consumption of the metal by the customer. One of the ways of reaching this aim is the reduction of consumption coefficients. The experience showed that 1.15 t of molten metal is required for 1 t of the rolled metal. Increase in output of efficient metal is the main principle of the ferrous metallurgy progress.

Before 1950 the metal casting at the metallurgical plants was performed exclusively into ingots. Total amount of ingot discard reached 20–25 % [1].

Metallurgists concentrated their efforts to decrease the discard by application of warming inserts, change in sizes of ingots, improvement of conditions of casting and others. But all these methods did not allow radical changing in existing situation.

During the 1950s a method of a continuous casting of steel (CCS) was developed and implemented successfully in the former USSR.

Highly-effective and highly-economic process of CCS was widely spread in the 1960–1970s at the enterprises of ferrous metallurgy of capital and industrially developing countries [1].

In 1970, 4.3 % of all steel was produced in the world using the machines of the continuous casting. During the period of 1970–1985 this volume increased in the industrialized and developing countries by 13 times, and by the middle of the 1990s the implementation of the continuous casting in the industrialized countries was almost finished [2].

At the present time, when all leading steel producers have passed to its continuous casting and large experience of its use has been gained, it is possible to conclude that hopes, set on this technology, were justified completely. Moreover, the new opportunities of an abrupt increase in effectiveness of all the ferrous metallurgy were open up. Updating of equipment and technology of CCS is continued. A great attention is paid to automation and computerization, that makes it possible to speed up the process of casting, to improve the quality of the billet produced [3].

It should be noted that application of CCS promoted a noticeable break in the ferrous metallurgy: all drawbacks in steel casting into ingots were eliminated, there is no need in blooming and slabbing mills in the cycle of a rolling processing.

At the same time, the problem of a power and economic nature, connected with a need in heating of slabs and their next rolling processing for sheet and sections, became unsolved.

Large technical-economic advantages of installation for CCS (ICCS), as compared to traditional metallurgical technology, promoted the rapid increase in capacities of this equipment in the industrialized countries.

In the period of 1971–2006 they increased in the USA approximately by 18 times, in Japan — by 14 times, in Germany — by 11 times.

By generalizing the technological and organizational-economic results of use of CCS, they can be divided into two following groups:

- total increase in effectiveness of production of metal products as a result of an abrupt reduction in

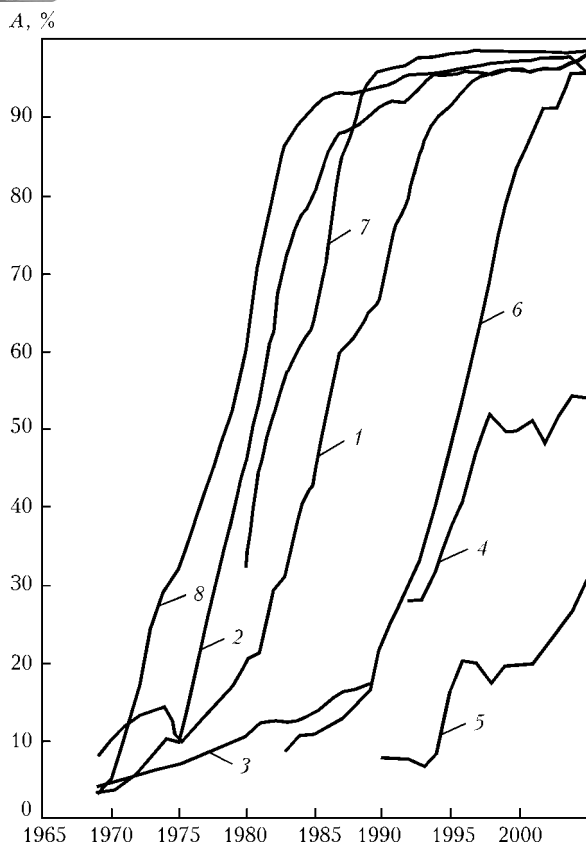


Figure 1. Change in share of steel A produced by CCS method in the total volume of production: 1 — USA; 2 — Germany; 3 — USSR; 4 — Russia; 5 — Ukraine; 6 — China; 7 — South Korea; 8 — Japan

consumption of molten steel per unit of ready rolled stock and noticeable increase in quality of ingots, decrease in capital investments for construction of shops and service expenses;

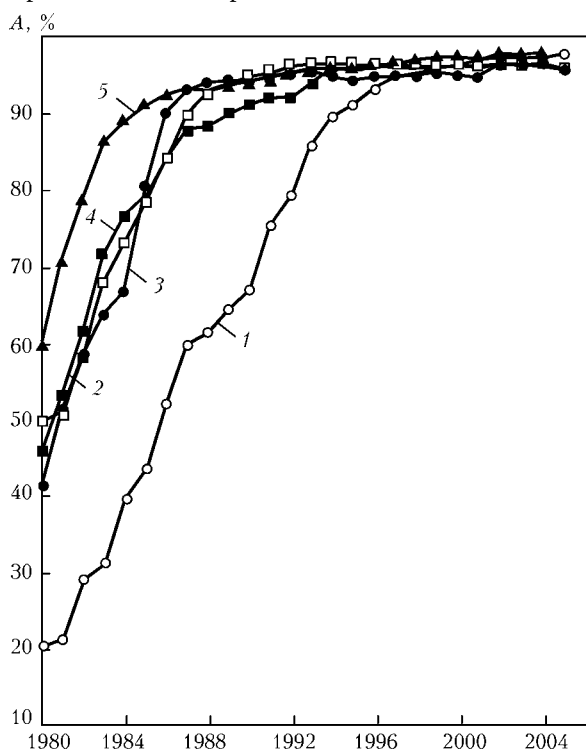


Figure 2. Share of steel A produced by CCS method in IC: 1 — USA; 2 — Germany; 3 — France; 4 — Italy; 5 — Japan

- feasibility of creation of a radically new technological complex from ICCS and rolling mills, which will change all the rolling manufacturing, reduce abruptly the volumes of necessary capital investments, liquidate the most expensive and metal-intensive equipment, reduce the service expenses.

Data about the progress of CCS by groups of countries are given in Figures 1–4. The share of CCS in all the industrialized countries (IC) (Figure 2), except USA, exceeded 90 % in 1990 and was more than 95 % by 2006.

Figure 3 shows the change in CCS share in countries with a planned and transition economy. Here, the noticeable lag in the development of the above-mentioned method, as compared with IC, should be noted. Only in Slovakia the share of CCS increased up to 100 % from 1990 until 1996. The continuous casting is developed successfully in Poland and China, its share by 1998 was about 70 %. In the former USSR the mentioned method of production was developed very slowly (from 12 % in 1980 up to 18 % in 1988). But after disintegration of the former USSR, Russia activated this process and the share of CCS reached 50 % by 1990.

The development of CCS is much the worse in Ukraine, where its share varies in the ranges of 30 % [4].

Development of CCS method in developing countries with a market economy is shown in Figure 4. Let us outline the following data:

- the share of CCS in South Korea reached 98 %;
- CCS in Brazil is developed intensively and reached 100 % by 2005;
- this index in the world approached 100 % by 2005;
- among the mentioned group of countries, the lowest level of CCS was observed in India, but it also exceeded 50 % in 1998.

CCS, replaced the casting into moulds, became the technological transition that provided the highest economic effect, the output of ready products was increased by 15 %.

Active scientific-technical development, industrial application of this highly-effective technology began in 1970s, and by 1990s almost all steel in IC was cast by using this technology. Even China, beginning to use CCS only in the 1980s, almost all steel subjected to casting using this technology, already by 2000.

In the former USSR the development of CCS began with a delay by 15–20 years and in 1990s it was stopped in general. Now, the share of CCS in Russia is about 54 %, in Ukraine — approximately 30 %. The cause of this situation consists in serious errors of management of this branch, planning bodies and central research institutions, which cannot be justified by a lack of funds and capital investments, as the economic losses have already exceeded much the required expenses.

The given analysis showed the following:

- technological structure of our ferrous metallurgy is lagging from industrialized countries by 20–25 years, that is inadmissible;



- to come out from the created situation, the immediate development and approval of governmental program of scientific-technical development of ferrous metallurgy of the country for the nearest five-ten years is required.

The beginning of the 1970s is characterized by a remarkable industrial implementation of machines for continuous casting of slab billets. Low-speed vertical machines for continuous casting of billets (MCCB) were replaced by radial and curvilinear machines with a much higher speed of casting. Main directions in progress of slab MCCB consisted in a creation of highly-efficient machines, allowing synchronizing operation of large oxygen-converter shops. As a matter of fact, this stage was finished in the middle of the 1980s by a practical embodiment of conception of high-efficient converter shops (capacity of converter above 300 t) with 100 % CCS. At present these shops are functioning successfully in all industrially developed countries of the world [5]. Simultaneously, technology of steel casting of super-wide slabs (2500–2600 mm) was realized. Transition from casting into ingots to CCS promoted also the more strict requirements to the quality of products.

Technology of CCS for slabs came into the phase of updating separate technological elements and optimizing functioning of MCCB with all the production cycle of the metallurgical plant. Here, the development of processes of continuous casting of slabs occurs under the conditions of a severe competition at the world market, the main requirements of which are the increase in quality of products at simultaneous reduction in their cost. At present, about 60 % of billets, cast by the method of a continuous casting, are produced in slab MCCB. In the world, the main manufacturers of continuously-cast slabs are Japan, USA, China, Germany, Korea and Russia. Their share is more than two thirds of the world volume of slab production. Now, there are more than 500 slab MCCB in the world with a total number of passages above 700 pieces [5].

MCCB for producing a «classic» slab are subjected during recently to serious design changes, which consist in the following:

- preferable are the MCCB with a vertical mould providing the increase in billet quality at a simultaneous 1.4–1.5 times increase in efficiency;
- curvilinear scheme of technological line of MCCB with a multi-point bending and unbending is used;
- feasibility of varying the billet width during casting is envisaged;
- capacity of industrial ladle (40–50 t) is increased, system of partitions is used for metal flow control;
- continuous measurement of metal temperature in industrial ladle and in separate zones of billet movement became obligatory;
- application of technology of metal preheating directly in an industrial ladle is widened;
- a part of operations for final making of steel is transferred to a ladle (argon blowing, treatment by a powdered wire);

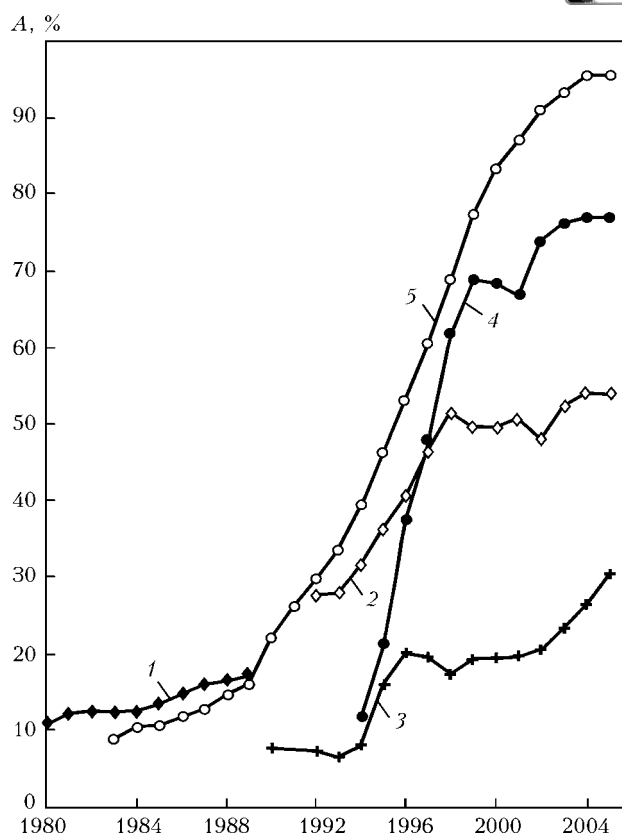


Figure 3. Share of steel A produced by CCS method in countries with a planned and transition economy: 1 — USSR; 2 — Russia; 3 — Ukraine; 4 — Poland; 5 — China

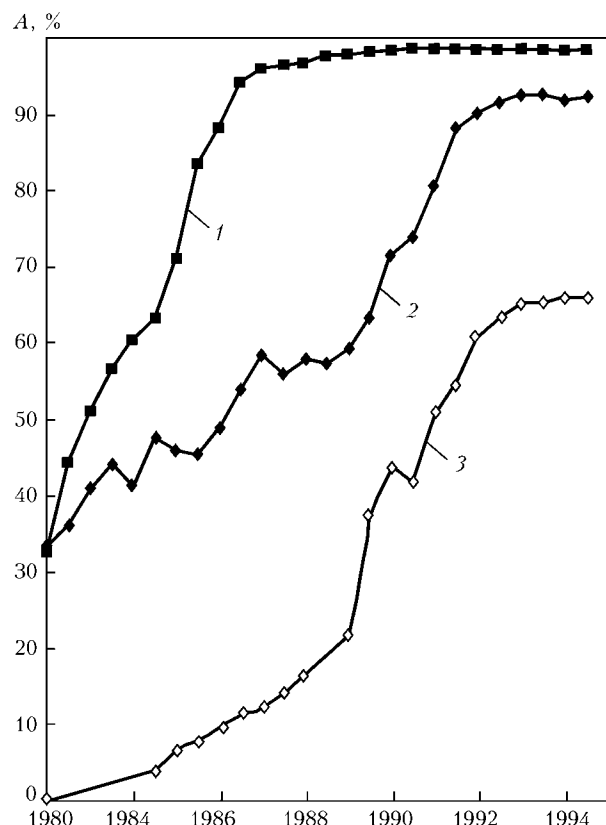


Figure 4. Share of steel A produced by CCS method in countries with a developing economy: 1 — South Korea; 2 — Brazil; 3 — India



• requirements became more strict to steel protection from secondary oxidation at the entire stage of steel transfer from a ladle to the mould;

• technology of a «soft» reduction of metal in the zone of a secondary cooling (ZSC) finds the wider spreading;

• electromagnetic stirring of metal below the mould is used;

• diameter of support rollers in ZSC is decreased;

• system of automatic control of metal level in mould, and also system of automatic feeding of slag mixtures into the mould is used (obligatory).

As a result of the mentioned changes the efficiency of MCCB was increased greatly, the quality of surface and internal structure of a slab was improved, expenses for manufacture of machine, its maintenance were reduced, etc.

Abrupt change in idea about the modern slab MCCB promoted the fact that most of machines in countries of Europe, Japan and the USA were subjected to a serious reconstruction and modification, caused mainly by characteristics attained in practice of an oxygen-converter process. First of all, it concerns mean daily and mean monthly efficiency of the modern converters (growth on average by 25–30 %). In this case, a simple increase in speed of a billet withdrawal in already available MCCB is seemed, as a rule, a rather complicated procedure with the need in increase of extension of zone of the billet cutting. From the other hand, the tendency of development of a world market of metal products proves that all preference is given to a slab billet of 150–200 mm thickness at its width of 1000–1200 mm. For a larger amount of MCCB the fulfillment of these requirements leads to the decrease in a mean efficiency of the machine as compared with a design efficiency.

Under such conditions, an important factor, increasing a mean monthly efficiency of MCCB is the increase in a series of casting between stops. As the practice of leading metallurgical enterprises of the USA and Japan shows, the duration of operation of

MCCB can be month and more at definite conditions (synchronizing of shop work and intermediate maintenance of MCCB). Undoubtedly, these characteristics can be considered as record. However, the routine practice of continuous casting of slabs is based even today on series of several days. Nevertheless, it should be noted that this technology requires radically new approaches to the automation of the casting process, steel preparation for casting at a ladle treatment, rational design of an industrial ladle and refractories for steel casting, etc.

Efficiency of MCCB, manufactured on the above principles, is 1.5–1.8 times higher than in MCCB, manufactured 15 years ago. It allows us to suppose that a great growth in a number of MCCB in the world will not grow, as their specific efficiency will increase. Some increase will be expected at the expense of countries of East Europe, Ukraine, Russia and Far East (first of all of China).

Among 11 large metallurgical plants of Ukraine, only five of them are equipped with ICCS to a larger or smaller extent (Table). Their main part is in service for long time and requires reconstruction. Moreover, only MCCB of the Iljich Mariupol Metallurgical Works are operating at exceeding of their design capacity. Most machines are used not in a full volume that is one of causes of increased energy consumption per ton of steel.

In spite of a comparatively low fraction of steel casting in MCCB, Ukraine has a great experience in producing billets of various sections. There is a number of achievements in a technological aspect, which are not inferior to the world characteristics.

In 1960 an industrial MCCB was put into service at the open-hearth shop of Donetsk Metallurgical Works for producing slabs, serving a prototype of many pioneering technical and design solutions used at many plants of the world. In 1984 a vertical four-passage blooming MCCB was put into service at the Experimental-Production Center (EPC) of Donetsk Metallurgical Works.

Main data on MCCB in Ukraine

Works (shop)	Volume of ladle, t	Type of ladle treatment	Year of start	Quantity of machines and passages	Billet section, mm	Efficiency project / 2001, mln t
Donetsk Metallurgical Works (open-hearth)	150	—	1960	1 × 4	200 × 1200 150 × 600	0.31 / 0.29
Metallurgical Works «Azovstal» (OCS)	350	UDM1	1977	3 × 2	200 × 1200 315 × 1900	3.7 / 2.8
«Drill Collars & Kellies Plant», Sumy (EPS)	18	ASEA-SKF VOD	1987	1 × 2	200 × 500	0.1 / 0.07
Iljich Metallurgical Works (OCS)	160	UDM	1993	2 × 1	250 × 1500 300 × 1900	1.5 / 2.4
Dneprovsky Metallurgical Works (OCS)	250	UDM	1994	2 × 6	335 × 400	1.5 / 0.82
CJSC MMW–ISTIL, Ukraine (EPS)	120	LF VOD	1999	1 × 6	100 × 150 120 × 180	0.8 / 0.3
In total:						7.91 / 6.68



In 1997 at Mariupol Metallurgical Works «Azovstal» one of the world largest oxygen-converter shops (2×350 t) with 100 % steel casting in slab MCCB was put into service. Maximum size of slab was 315×1900 mm. In 1993 the oxygen-converter shop (OCS) of the Iljich Mariupol Metallurgical Works, having three converters of 160 t capacity and two single-passage MCCB with 315×1900 mm maximum section of slab, was reconstructed.

In 1994 the oxygen-converter shop of Dneprovsky Metallurgical Works, having two converters of 250 t capacity and two six-passage blooming MCCB, was reconstructed. One machine was reconstructed to produce 160×160 mm square. Wide range of sections of a blooming billet (from 200×200 mm up to 500×500 mm) is produced by MCCB of plant «Drill Collars & Kellies Plant» (Sumy). At this plant a large amount of alloyed and high-alloy quality grades of steels are subjected to casting.

In total, 12 MCCB units are operating in Ukraine, seven of them are used for casting slab billets and five of them — for section billets. In 2005, 29.5 % of steel was cast by them [6].

In 1996 the company «Metalsrussia», incorporated into the company ISTIL Group (USA), purchased a controlling block of shares of Donetsk Metallurgical Works. This allowed company to select several shops at the plant and to organize the new enterprise MMW ISTIL (Ukraine). More than 85 million dollars were invested that allowed creation of an updated complex of a ladle treatment and CCS, which was manufactured and put into service in 1999. Complex includes an advanced 120-ton electric arc furnace, furnace-ladle, vacuum station, six-passage MCCB. The later allows producing a section billet of square and round sections.

In 2002 the realization of the first stage of reconstruction of oxygen-converter shop of Enakievo Metallurgical Works was finished, resulting in creating a flexible technological module (ladle-furnace and six-passage high-speed section MCCB), combined with operation of converter of 160 t capacity. All equipment for furnace-ladle aggregate and MCCB was manufactured by the Ukrainian manufacturer — Novokramatorsk Machine-Building Works.

Specifics of reequipping of metallurgical enterprises of Ukraine by machines of CCS consists in the following:

- metallurgical enterprises have available mainly the steelmaking units of a high unit power that supposes the application of high-efficient slab and section MCCB;

- in operating steelmaking shops a general project does not envisage, as a rule, a casting span for arrangement of MCCB that requires the construction of infrastructure, required for their functioning in a strictly limited space;

- reconstruction of steelmaking production is preferable without its stop to avoid losses in volumes of production;

- it is necessary to include obligatory a system of technological measures for improving the quality of

molten steel that supposes the presence of an integrated ladle treatment of metal;

- it is rational to use as much as possible the capabilities of the Ukrainian machine-building plants, which have already gained a definite experience in design and manufacture of ladle-furnace aggregates and MCCB of different types.

According to estimates of Ukrainian experts, to have an effective restructuring of steelmaking shops in Ukraine for 10 years, it is necessary to construct 8-9 multi-passage section MCCB, 2-3 slab MCCB, as well as 5-6 units for ladle steel treatment of a furnace-ladle type. In addition, a serious reconstruction is required for already functioning MCCB [7].

Main technical-economical trends of CCS development on the basis of the world experience. The most serious scientific break seems to be a creation of MCCB for casting slabs of a mean thickness by a number of Western companies, providing the sheet quality at the level of a traditional slab. Advantages of this technology is the high quality of products, feasibility of producing slabs of width of more than 2500 mm, reduction in losses of energy and metal during the next processing, etc.

In this case the technology of producing a sheet was based on conception of casting-rolling modules [8]. This created necessary conditions for the construction of mini-plants, compatible by a number of parameters with integrated plants working by traditional technology. Casting-rolling modules, developed on the base of CCS for slab of a medium thicknesses, find at present the more and more wide spreading in the developing countries (where need in a sheet rolled metal is increased). Owing to use of most advanced achievements in the field of CCS (electromagnetic stirring, «soft» reduction, immersed sleeve of «beaver's tail» type, etc.), these modules will continue their progressing also in future [9, 10].

One of the directions of improving slab MCCB is to satisfy the growing requirements for the quality of products. For these purposes, the updated straight-linear moulds, which can be used also in already operating machines after appropriate reconstruction, are more often used in practice. As a whole, the advanced developments in the field of designs of moulds are focused on the following:

- increase in casting speed by intensification of conditions of cooling and frequency of swinging, optimizing of law of the mould oscillations;

- effect on metal flows in mould by electromagnetic braking for reduction of a single-sided assimilation of non-metallic inclusions in use of a curvilinear mould, and also for smoothing fluctuation of pool level at very high speeds of casting;

- further increase in life of plates of the mould using new technology of deposition of coatings, etc.

The priority trends in updating design of MCCB include application of a hydraulic drive to provide the mould swinging using a non-sinusoidal law of oscillations. Optimizing of motion of steel flows in mould molten pool has a great importance from the point of view of the slab quality improvement. A



special attention is paid to the optimizing the geometric shape of immersed sleeves. Application of new materials and methods of treatment of refractory products makes it possible to increase the resistance of immersed sleeves against thermal shocks, to decrease the wear rate in the zone of a slag band and intensive motion of molten steel flows, as well as to prevent the clogging of inner cavities in casting of steels deoxidized by aluminium.

At present, the method of electromagnetic braking of steel flow, entering the mould, finds the wide spreading. This gives an opportunity to decrease greatly the speed of motion of flows, to limit their penetration inside the liquid phase of billet and also to provide their rational motion.

With increase in quality of internal zones of continuously-cast ingots, a special attention is paid to the conditions of billet movement in ZSC. Except the objective processes of liquation development during solidification, the intensity of cooling, ferrostatic pressure, keeping of geometric sizes, depending on arrangement of guiding rollers, etc., influence greatly the quality of the billet. It is understandable, that it is very difficult in practice to eliminate these factors without additional measures.

From information of some researchers, this problem is rather successfully solved in use of the method of a «soft» reduction of the billet at the end of the molten zone. The main positive effects of the above-mentioned method consist in the following: a great suppression of axial porosity is occurred, coefficient of axial liquation is reduced and conditions of formation of crystalline structure in the zone of equiaxial crystals are improved, V-shaped liquation is decreased.

At the end of the 1980s and beginning of the 1990s a new direction of a continuous casting for slabs, i.e. casting of thin slabs in MCCB, combined with a rolling mill, was formed in the world. As any new process, the casting of thin slabs has a number of promising peculiarities, but it is also characterized by some «narrow» places.

The recent results of industrial application of method of casting of thin slabs showed that the constructive realization of these MCCB requires radically new solutions relative to the following problems: design of the mould, optimizing of supply of molten steel from an industrial ladle into mould, using method of «soft» reduction and special system for prevention of breaks, development of a special system of a scale removal, optimizing of negative time of ingot stripping, etc.

At the same time, it is possible to state today that technology of continuous casting of thin slabs with reduction in the presence of liquid phase with a subsequent rolling makes it possible to decrease the capital expenses and expenses for processing, to improve the properties of ready products at uniform quality of the surface.

Thus, it is possible to formulate the following general tendencies in development of CCS process using slab MCCB:

- in the coming years there will no great quantitative increase in volume of continuously-cast slabs;

main attention will be paid to the problems of quality of billets, reduction of expenses for casting, etc.;

- improvement of quality of surface of continuously-cast slabs will be attained by updating the design of moulds (straight-linear mould) using a hydraulic drive for mechanism of swinging;

- parameters of metal flow in mould will be improved as a result of use of immersed sleeves of optimum geometric shape and method of electromagnetic stirring;

- development of functional capabilities of industrial ladle will occur due to increase of their service characteristics;

- systems of a progressive multi-point bending and unbending of billet will be used for decrease its damage by internal cracks;

- method of «soft» reduction of billet with a molten middle to suppress the axial porosity and liquation will find application;

- conceptions of continuous casting of thin slabs with their reduction and next rolling in hot state and appropriate increase in competitiveness of mini-plants at the world market of sheet products will be developed;

- degree of automation of operation of MCCB will be increased with including into a single system of all parameters of molten steel, conditions of steel cooling in the mould and zone of the secondary cooling in the presence of a feedback.

Urgent technical-economic and organizational measures for the CCS progress in metallurgy of Ukraine. CSS is the largest technical achievement of the metallurgy. Application of this process in the production allows saving up to 20 % of metal and reducing greatly the metallurgical cycle by elimination of reduction mills from the process. Share of steel casting in ICCS approached in most countries of the world to limiting one and is equal to 95–97 % [11].

It should be noted that this value in Ukraine is not more than 29 % that is a certain achievement, because five years ago it was twice lower.

As a result of analysis we consider that there is a unique feasibility at present for the Ukrainian enterprises, having no complexes of CCS, to use the most advanced technological schemes, allowing improving the metal quality and decreasing its power consumption in casting in ICCS.

CONCLUSIONS

1. It is necessary to decrease the section of billets being cast with increase in speed of steel casting.

2. Vertical machines with bending of billets into horizontal position should be used.

3. It is necessary to use industrial ladles of increased capacity and height, moulds with height-variable conicity at the more perfect condition of swinging, braking of steel flows in the mould.

4. It was established that to decrease the damage of the billet by internal cracks, it is necessary to use systems of progressive multi-point its bending and unbending.



5. It was defined, that to suppress the axial porosity and liquation, it is necessary to use the method of «soft» reduction of the billet with a molten middle.

6. It is necessary to develop conceptions of a continuous casting of thin slabs with their reduction and next rolling in hot state and appropriate increase in competitiveness of mini-plants at the world market of sheet products.

7. It is necessary to increase the degree of automation of MCCB operation with inclusion into a single system of all the parameters of molten steel, conditions of cooling of steel in the mould and ZSC in the presence of a feedback.

1. Myrtyymov, F.M. (1978) *Ferrous metallurgy of capitalist and developing countries*. Kiev: Naukova Dumka.
2. Medovar, B.I. (1990) *Metallurgy yesterday, today, tomorrow*. Kiev: Naukova Dumka.

3. Efimenko, G.G., Neshchadim, V.N., Tsymbal, M.I. et al. (1993) *Analysis of development and technical-economical problems of steelmaking progress in world*. Dnepropetrovsk: DMetI.
4. Najdek, V.L. (2006) Address to participants of meeting MNTS. *Metall i Litiyo Ukrainy*, **1**, 5–6.
5. Smirnov, A.N., Pilyushenko, V.L., Minaev, A.A. et al. (2002) *Processes of continuous pouring*. Donetsk: DonNTU.
6. Kharakhulakh, V.S. (2006) State-of-the-art of steelmaking in Ukraine and prospects of development with introduction of current technologies up to 2006. *Metall i Litiyo Ukrainy*, **1**, 4.
7. Malik, A.A., Orman, V.Ya., Dovgach, V.Yu. (2006) Facilities of steelmaking shops of Ukraine with out-of-furnace units and continuous melting units. *Ibid.*, **1**, 12.
8. Ganzhin, V., Kiselyov, Yu. (2003) Technology of 21st century as the prospects of Russia. *Nats. Metallurgiya*, **1**, 77.
9. Smirnov, A.N. (2006) Current tendencies of development of steel pouring processes. *Metall i Litiyo Ukrainy*, **1**, 7–11.
10. Matveev, B.N. (2005) Production of thin hot-rolled bands in LPA ISP (the present and future). *Elektrometallurgiya*, **5**, 26.
11. Belyanchikov, L.N., Borodin, D.I., Valavin, V.S. et al. (2001) *Steel at the boundary of centuries*. Moscow: MISiS.

UNIVERSAL MACHINE FOR ELECTRON BEAM WELDING MODEL 102



- Universal production machine for EBW of a wide range of cylindrical and flat workpieces.
- PC and programmable controllers are used.
- Real-time seam tracking and monitoring of EBW process by RASTR-3 system on the basis of secondary electron emission.
- Power source with the electron tube flashless system.
- Mobile type 15, 30 or 60 kW electron beam gun at 60 kV.

Machine design

The work chamber has two sliding doors. For loading and unloading, the workpiece table is moved out of the work chamber onto the runout platform. This is especially necessary for welding of large heavy workpieces with commensurately large clamping devices. The table accommodates the universal rotator with horizontal or vertical axis, and back centre. The EB gun 3-axis-manipulator has the travelling distance in X-direction — 2000 mm, in Y-direction — 800 mm and in Z-direction — 800 mm. Precision of the guidance and drive systems equals that of the precision machine tools operation with tolerances in the hundredth-of-a-millimeter range.

The gun can be mounted in any spatial position and has an independent turbomolecular pumping system. The cathode area is isolated by a vacuum valve to keep the gun under vacuum when the work chamber is vented.

Application. Welding and repair in turbine engine fabrication; selection of welding technology for a wide range of workpieces.

Contacts: Prof. Nazarenko O.K.
E-mail: nazarenko@technobeam.com.ua



COMPARATIVE EXPERIMENTAL-INDUSTRIAL INVESTIGATIONS OF EFFECT OF 65 % FERROSILICIUM WITH DIFFERENT CONTENT OF CALCIUM ON CONTAMINATION OF STEEL ShKh15SG-V WITH GLOBULAR ALUMOCALCIUM INCLUSIONS*

A.I. PANCHENKO¹, A.S. SALNIKOV¹, I.N. LOGOZINSKY¹, L.N. KOROL¹, S.S. KAZAKOV¹,
M.I. GASIK² and A.P. GOROBETZ²

¹OJSC «Dneprospetsstal», Zaporozhie, Ukraine

²National Metallurgical Academy of Ukraine, Dnepropetrovsk, Ukraine

Results of comparative investigations of effect of 65 % ferrosilicium with different content of calcium on contamination of steel ShKh15SG-V with globular alumocalcium inclusions were summarized and analyzed. It was established that the application of imported 65 % ferrosilicium with 0.08 % Ca reduces the calcium content in steel as compared with a filler of ferrosilicium of grade FS65 with 0.42 % Ca (DSTU 4127-02) and decreases the amount of globular inclusions in billets and ready metal products. However, the contamination of metal with oxide inclusions is increased in this case.

Key words: steel ShKh15SG-V, electric melting, basic process, ferrosilicium, deoxidation, slag, silicon, calcium, aluminium, vacuum treatment, methods of evaluation of non-metallic inclusions, globules, oxides, sulphides

One of the main characteristics of quality of steels of class ShKh at acceptance inspection is the degree of contamination of metal with endogeneous hard-to-form (globular) inclusions of a complex composition. Assessment of quality of metal products is made by the methods of optic microscopy by a maximum point of contamination on scales of GOST 801-78 for consumers of domestic market and ASTM E-45 (method A), DIN 50602 (method K) for consumers of the foreign market. Characteristics of crystalline-chemical structure and dimensional factor of inclusions define in this case the quality and effectiveness of technology of melting steels ShKh as a whole.

Under conditions of «Dneprospetsstal» Works the steels of ShKh15 and ShKh15SG-V grades are melted in basic arc electric furnaces DSP-50 with next alloying by carbon ferromanganese (DSTU 3547-97), ferrosilicium FS65 (DSTU 4127-2002), deoxidation and desulphurization in installation «furnace-ladle» (IFL) under slag melt of CaO-CaF₂ system. Final deoxidation by aluminium, correction of chemical composition and metal degassing are realized at the stage of a ladle vacuum treatment. Duration of metal treatment in IFL is 40-50 min, vacuum treatment is not less than 12-15 min. Steel is poured to mould into 3.6-4.8 t mass ingots using a jet protection with argon. According to data, obtained during recent years, the output of efficient metal after the first

acceptance inspection by non-metallic inclusions did not exceed 60-70 % for steel ShKh15SG-V and 80-95 % for steel ShKh15. From results of acceptance inspection in 2006, the fraction of rejections by groups of non-metallic inclusions distributed as follows, %: globules 55.5, oxides 31.2, sulphides 13.9. Using methods of metallography the alumocalcium compounds of CaO-Al₂O₃ with fringing of separate inclusions by sulphide phase CaS were determined in composition of globular inclusions.

The relation was established between mass fraction of calcium in metal and amount of rejections of acceptance inspection by inclusions. Thus, at calcium content at the level of 6.5, 9.2 and 11 ppm the fraction of rejections by globular inclusions was, respectively, 9, 13.5 and 17.5 % (Figure 1). In this case, the mass fraction of calcium in steel ShKh15SG is on average by 30 % higher than the similar characteristic for steel ShKh15 (Figure 2).

Parallel with characteristics of metal quality by non-metallic inclusions, the residual content of oxygen of not more than 10-20 ppm is strictly regulated in it, and by the requirements of separate foreign consumers the sulphur is regulated of not more than 15 ppm.

Continuously growing requirements of the international market of metal products to the quality of steel of class ShKh15 stipulate the need in carrying out of regular investigations both on stabilization of technological parameters of a through technology of melting and ladle treatment of metal, and also on control of processes of formation of composition and amount of inclusions. Regulated composition of inclu-

*In this work A.E. Koval, P.A. Shebeko, L.M. Skripka and A.V. Zhajvoronok took part.

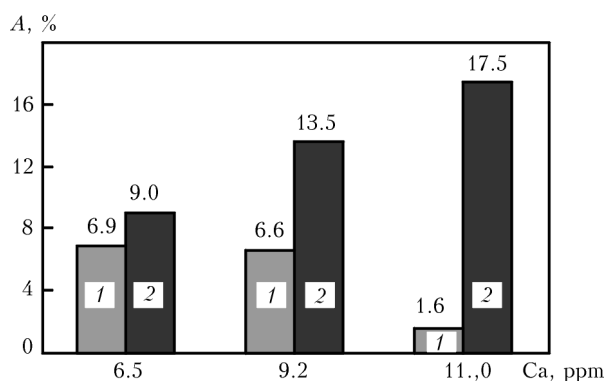


Figure 1. Contamination of steel ShKh15SG-V by non-metallic inclusions depending on calcium content: 1 — oxides; 2 — globules; A — rejections

sions is attained by control of content and (more important) by regulation of ratio of activity of elements dissolved in metal, such as calcium, aluminium, silicon, oxygen, sulphur, and magnesium in some cases [1].

As it follows from the given data of acceptance inspection and metallographic analysis of inclusions, element, participating actively in the formation of oxide and sulphide phases is calcium. Calcium solubility in iron-carbon melts is higher than that in pure iron. Grade content of carbon in steel ShKh15SG is 0.95–1.05 %, therefore, probably, the equilibrium content of calcium in molten steel, providing other conditions equal, is higher than in melt of pure iron.

Using existing technology of melting ShKh15SG steel, one of the main sources of calcium enter to steel is domestic high-silicon ferrosilicium with 0.5 % Ca, whose content, unfortunately, is not regulated either by Interstate standard GOST 1415–93, or DSTU 4127–2002, or ISO 5445:1980.

It should specially noted that the second (and by the degree of effect can be also the first) source of calcium enter into the metal is a slag melt of CaO–CaF₂ system due to calcium reduction by silicon in metal deoxidation by ferrosilicium and aluminium, dissolved in molten steel. When working out the technological instructions on out-of-furnace steel refining in furnace-ladle and analysis of calcium enter, this circumstance is not taken into account. At the same time, silicon of ferrosilicium (as also silicon assimilated by steel from ferrosilicium) interacts with CaO of slag melt by the following scheme:

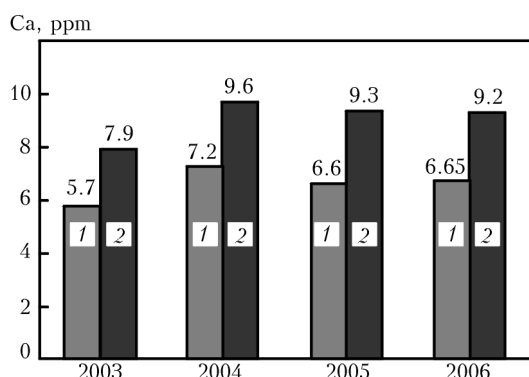


Figure 2. Calcium content in steel ShKh15 (1) and ShKh15SG (2) depending on year of production

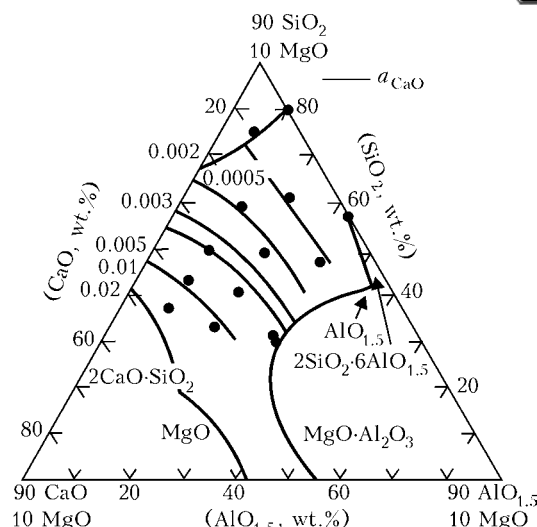
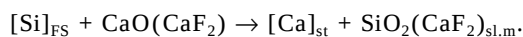
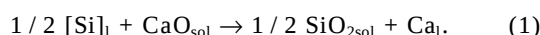


Figure 3. Activity of CaO in CaO–Al₂O₃–SiO₂ pseudo-ternary system at 10 % MgO and temperature 1550 °C [2]



Assessment of degree of calcium reduction by silicon is based on thermodynamic analysis of reaction



Relationship of change in Gibbs energy by reaction (1) is presented by the following expression [4]:

$$\Delta G^\circ = 156000 - 20T \text{ [J / mol]}.$$

Consequently, analytic expression of constant of equilibrium of reaction (1) is determined as

$$K_1 = \frac{\alpha_{\text{SiO}_2}^{1/2} \alpha_{\text{Ca}}}{\alpha_{\text{Si}_2}^{1/2} \alpha_{\text{CaO}}} = \frac{(\gamma_{\text{SiO}_2}^{1/2} X_{\text{SiO}_2}^{1/2}) (\gamma_{\text{Ca}} X_{\text{Ca}})}{(\gamma_{\text{Si}_2}^{1/2} X_{\text{Si}_2}^{1/2}) (\gamma_{\text{CaO}} X_{\text{CaO}})} \quad (2)$$

or in a numerical representation as

$$\lg K_1 = -\frac{8144}{T} + 1.044. \quad (3)$$

Activity CaO in four-component system CaO–MgO–Al₂O₃–SiO₂ at 10 % MgO (Figure 3) is 0.02 and, taking activity of SiO₂ with account for compositions of slags close to 0.001 (Figure 4), the expression (3) acquires the following form:

$$\lg K_1 = \frac{(0.001)^{1/2} \alpha_{\text{Ca}}}{\alpha_{\text{Si}}^{1/2} 0.02} = 1.571 \frac{\alpha_{\text{Ca}}}{\alpha_{\text{Si}}^{1/2}}. \quad (4)$$

Assuming the solutions of silicon and calcium in iron-carbon melt in steel ShKh15SG-V as finitely di-

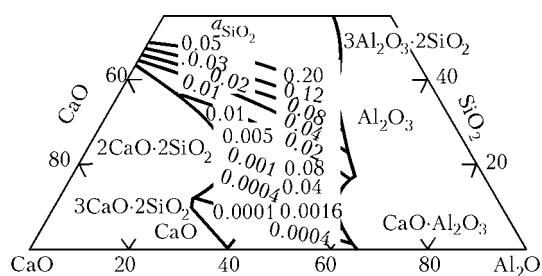


Figure 4. Activity of SiO₂ in CaO–Al₂O₃–SiO₂ system at temperature 1600 °C [3]

**Table 1.** Chemical composition of metal at four stages of technology of steel ShKh15SG-V production

Period of melting	Element content, %							
	Ñ	Mn	Si	Cr	Đ	S	Al	Ñà
Tapping from DSP (I)	$\frac{0.75}{0.033}$	$\frac{0.15}{0.0016}$	$\frac{0.13}{0.0027}$	$\frac{1.2}{0.0125}$	$\frac{11 \cdot 10^{-3}}{1.6 \cdot 10^{-4}}$	$\frac{4.5 \cdot 10^{-3}}{5 \cdot 10^{-4}}$	$\frac{4 \cdot 10^{-3}}{7 \cdot 10^{-5}}$	$\frac{4 \cdot 10^{-4}}{5 \cdot 10^{-6}}$
Start of treatment in IFL (II)	$\frac{0.95}{0.042}$	$\frac{0.17}{0.0021}$	$\frac{0.20}{0.0038}$	$\frac{1.2}{0.0125}$	$\frac{14 \cdot 10^{-3}}{1.9 \cdot 10^{-4}}$	$\frac{18 \cdot 10^{-3}}{2.7 \cdot 10^{-4}}$	$\frac{40 \cdot 10^{-3}}{12 \cdot 10^{-5}}$	$\frac{5 \cdot 10^{-4}}{5.4 \cdot 10^{-6}}$
End of treatment in IFL (III)	$\frac{0.99}{0.044}$	$\frac{1.0}{0.0097}$	$\frac{0.48}{0.0091}$	$\frac{1.43}{0.0148}$	$\frac{14 \cdot 10^{-3}}{1.6 \cdot 10^{-4}}$	$\frac{9 \cdot 10^{-3}}{1.6 \cdot 10^{-4}}$	$\frac{45 \cdot 10^{-3}}{91 \cdot 10^{-5}}$	$\frac{14 \cdot 10^{-4}}{18.7 \cdot 10^{-6}}$
After treatment in vacuumizer (IV)	$\frac{0.99}{0.044}$	$\frac{1.0}{0.0097}$	$\frac{0.48}{0.0091}$	$\frac{1.43}{0.0148}$	$\frac{14 \cdot 10^{-3}}{1.6 \cdot 10^{-4}}$	$\frac{5 \cdot 10^{-3}}{1 \cdot 10^{-4}}$	$\frac{33 \cdot 10^{-3}}{64 \cdot 10^{-5}}$	$\frac{11 \cdot 10^{-4}}{14.5 \cdot 10^{-6}}$

Note. 1. Fe — base. 2. In numerator wt.% are indicated, in denominator — mol %.

Table 2. Calcium content in metal during technological process by periods of melting, ppm

Method of determination	I	II	III	IV
Experiment	4	5	14	11
Calculation	15	18	27	21

luted, for which $\gamma_{Si}^{\infty} = 0.0013$, $\gamma_{Ca}^{\infty} = 2270$ [5], expression (4) is transformed into form

$$\lg K_1 = 1.571 \frac{2270 X_{Ca}}{(0.0013)^{1/2} X_{Si}^{1/2}} \quad \text{or} \quad \lg K_1 = 1.571 \frac{129.40 X_{Ca}}{X_{Si}^{1/2}} \quad (5)$$

and, consequently,

$$X_{Ca} = \frac{\lg K_1 X_{Si}^{1/2}}{129.40}. \quad (6)$$

Averaged values of content of elements at the stages of technological processing of steel ShKh15SG-V are given in Table 1.

Table 2 gives an actual content of calcium in metal during existing technology (periods I–IV) and calculated by formula (6) at temperature 1550 °C and values of constant of equilibrium $K_1 = 3.77 \cdot 10^{-4}$.

With a search for the solution of the problem of reduction of calcium amount in steel ShKh15SG-V by using imported ferrosilicium with a low content of calcium as compared with ferrosilicium supplied by domestic enterprises of Ukraine and plants of CIS, a comparative comprehensive analysis of chemical composition of presented samples of ferrosilicium was made (Table 3).

Peculiarities of technology of steel melting using imported ferrosilicium. In melting steel ShKh15SG-V, two technological variants, differed by method of adding imported ferrosilicium filler, were tested.

Melting by variant A. From the furnace with furnace slag cut-off a semi-product, non-deoxidized by silicon and aluminium was produced mainly. The ladle slag was formed by fillers of lime and fluorspar (1 % of metal mass) at CaO:CaF₂ ratio of 3:1. Further, the metal deoxidation by aluminium was made in ladle in the process of tapping (32–90 kg per melting, on average 1 kg / t). After aluminium, the ferrosilicium was added to the ladle from 200 up to 400 kg per melting (263 kg on average). Content of aluminium was corrected in IFL using a tribe-apparatus (6–19 kg) or aluminium pigs (from 16 up to 40 kg), and then ferrosilicium from 130 up to 300 kg (231 kg on average for melting) was added to IFL.

Melting by variant B. From furnace a semi-product was tapped, deoxidized only by 30–84 kg of aluminium per melting (38 kg on average) or 0.63 kg / t. Amount of aluminium, added to the ladle, was on average 55 kg per melting (0.9 kg / t). Aluminium fillers in IFL were on average 21 kg (0.35 kg / t). During metal vacuum treatment the metal was corrected by aluminium both by wire and also aluminium pigs on pins (7.7 kg on average for melting, 0.13 kg / t). Thus, using variant B, the metal tapped into ladle, had a higher deoxidation by aluminium as compared with variant A. Ferrosilicium was added mainly to the ladle during metal tapping from the furnace (523 kg on average per melting). Technological parameters of steel melting by variants A and B are given in Table 4.

Behaviour of elements in the process of steel melting. *Behaviour of silicon.* Content of silicon (Ta-

Table 3. Chemical composition of lumpy imported and domestic ferrosilicium

Ferrosilicium	Elements, wt. %								
	Si	Ñà	Ñ	Al	S	Đ	Ti	Pb	Mn
FS65 import certificate	65	0.08	0.04	0.08	0.010	0.020	0.034	0.001	0.20
FS65 imported (analysis OJSC «DSS»)	66	0.05	0.06	0.10	0.010	0.027	–	0.001	0.23
FS65 domestic	65	0.42	0.04	2.00	0.003	0.023	0.200	0.015	–

**Table 4.** Technological parameters of experimental melting of steel ShKh15SG-V using imported ferrosilicium

Variant	Number of meltings	Fillers, kg					
		FS65		Primary aluminium			
		Ladle	IFL	Furnace	Ladle	IFL	VD
À	8	$\frac{200-400}{263}$	$\frac{130-300}{231}$	—	$\frac{32-90}{61}$	$\frac{6-40}{24}$	$\frac{0-8}{1}$
Â	13	$\frac{400-550}{523}$	$\frac{0-70}{16}$	$\frac{30-96}{38}$	$\frac{45-84}{55}$	$\frac{16-32}{21}$	$\frac{4-21}{7.7}$

Note. Here and in Tables 5 and 7, the nominator shows the range of values, the denominator shows mean values; VD — vacuum degassing.

ble 5) in the first sample in IFL in melting by variant A was on average 0.26 %, by variant B — 0.50 %. Thus, silicon content in sample 1 in IFL by variant B of melting was at the medium-grade level (0.5 %), while in sample 2 it corresponded to ladle sample A by variants A and B of melting.

Behaviour of aluminium. Content of aluminium (Table 5) in sample 1 in IFL was on average 0.046 and 0.40 %, respectively, by variants A and B, while in sample 2 it was reduced on average to the level of 0.041 (by variant A) and to 0.034 % (by variant B) that was stipulated both by its oxidation, and also by removal from metal in the form of aluminium-containing oxide non-metallic inclusions.

During metal treatment in a ladle vacuumizer, the amount of aluminium reduced to 0.029 (variant A) and to 0.027 % (variant B) (Table 5).

Behaviour of calcium. Content of calcium was determined in samples of metal from furnace (3 samples), in processing in IFL (2 samples), vacuumizer (1 sample) and from ladle sample (Table 5). In sample 1 in IFL the amount of calcium is increased due to fillers of ferrosilicium into the ladle in metal tapping. In this case, the mean content is 4.9 ppm in variant A of melting, and 5.8 ppm in variant B. Increment of calcium by

variant B (1.3 ppm), as compared to variant A (0.6 ppm), is stipulated by increased fillers of ferrosilicium into ladle in metal tapping.

The largest difference in calcium content by variants of melting is observed in sample 2 in IFL. As is seen from Table 5, an intensive increment of calcium content from sample 1 to sample 2 with IFL by 12 ppm on average was observed by variant A of melting. In melting by variant B in IFL treatment the increment of calcium content was 2.4 ppm on average. Moreover, in meltings with ferrosilicium fillers in IFL the increment in calcium amount was 2.4 ppm on average, and without fillers — 1.0 ppm. At large filler contents by variant A of melting the increment was on average 12 ppm, and at small amounts of filler — 2.4 ppm. Thus, increase in calcium content in steel occurs mainly during treatment in IFL and depends on the amount of ferrosilicium added to IFL, that is, probably, associated with reduction of calcium by silicon from CaO of slag and confirmed by data of thermodynamic calculations.

During metal treatment in a vacuumizer and in the process of its casting the calcium content was reduced as compared with its level at the end of steel treatment in IFL. Thus, the calcium content in a ladle

Table 5. Change in content of elements in melting and ladle treatment of steel ShKh15SG-V in samples 1–3

Variant	Element content in samples during melting							
	Ñà, ppm						Si, %	
	Furnace			IFL		VD	IFL	
	1	2	3	1	2		1	2
À	$\frac{4-8}{5.8}$	$\frac{4-9}{5.4}$	$\frac{4-5}{4.3}$	$\frac{4-5}{4.9}$	$\frac{9-27}{17.0}$	—	$\frac{0.17-0.39}{0.26}$	$\frac{0.45-0.57}{0.50}$
Â	$\frac{3.8}{6.6}$	$\frac{3-9}{5.3}$	$\frac{3-7}{4.6}$	$\frac{3-1}{5.8}$	$\frac{5-9}{8.2}$	$\frac{6-1}{7.7}$	$\frac{0.41-0.57}{0.50}$	$\frac{0.46-0.57}{0.51}$

Table 5 (cont.)

Variant	Element content in samples during melting						
	VD	Al·10 ⁻³ , %			Ladle sample		
		IFL		VD	Ñà, ppm	Si, %	Al·10 ⁻³ , %
		1	2				
À	—	$\frac{36-58}{46}$	$\frac{32-55}{41}$	—	$\frac{9-13}{11}$	$\frac{0.50-0.56}{0.50}$	$\frac{24-36}{29}$
Â	—	$\frac{31-61}{40}$	$\frac{23-46}{34}$	$\frac{20-35}{27}$	$\frac{5.8}{6.5}$	$\frac{0.48-0.57}{0.51}$	$\frac{17-34}{27}$



sample in variant A of melting was 11 ppm (decrease by 6 ppm), and in variant B — 6.5 ppm (decrease by 2.7 ppm).

Evaluation of quality of experimental metal by non-metallic inclusions. Contamination of billet and ready products by non-metallic inclusions was determined by GOST 801–78 (Table 6) and ASTM E-45, method A (Table 7). Data of variants A and B of experimental meltings, as well as data about contamination of metal melted using the technology operating since November, 2006 until January, 2007, were compared (Table 8).

Evaluation of metal by GOST 801–78 (Table 8). As follows from data of Tables 7 and 8, output of efficient experimental metal (group 1) was 44.4 against 23.7 % in conventional technology of melting.

Metal of ready metal products of variant A is characterized by the increased contamination by globular inclusions, whose fraction from total amount of rejections (O + S + G) is 91.7 %. Number of samples with rejections by globular inclusions in experimental metal reaches 17.5 %, and using the operating technology — 13.6 % (Table 8). Total rejections by all types of inclusions (O + S + G) for experimental

Table 6. Contamination of experimental meltings of steel ShKh15SG using imported ferrosilicium by non-metallic inclusions at the primary evaluation by GOST 801–78 (acceptance inspection and research tests)

Purpose, group	Type of control	Variant of technology	Quantity, pcs			Type of non-metallic inclusions										Output of efficient metal, %	
			Melting	Batch	Sample	Î			S			G		Î + S – G			
						Mean value	Sample with rejection		Mean value	Sample with rejection		Mean value	Sample with rejection		pcs		%
							pcs	%		pcs	%		pcs	%			
Billet																	
1	Acceptance	Â	1	1	9	2.00	No data		2.00	No data		2.44	5	55.6	5	55.6	0
		Â	9	9	113	2.02	15	13.3	1.95	3	2.7	1.88	20	17.6	38	33.6	44.4
		Σ	10	10	122	2.03	15	12.3	1.97	3	2.5	2.16	25	20.5	43	35.2	40
	Research tests	Â	2	2	18	1.97	3	16.7	1.86	No data		1.94	3	16.7	6	33.3	0
		Σ	2	2	18	1.97	3	16.7	1.86	»	»	1.94	3	16.7	6	33.3	0
3	Acceptance	Â	1	1	15	2.33	1	6.7	2.50	»	»	2.20	1	6.7	2	13.3	0
		Σ	1	1	15	2.33	1	6.7	2.50	»	»	2.20	1	6.7	2	13.3	0
1+3	Acceptance + research tests	Â	2	2	24	No data	1	4.2	»	»	No data		6	25.0	7	29.2	0
		Â	11	11	131	»	18	13.7	»	3	2.3	»	23	17.5	44	33.5	44.4
Ready metal products																	
1	Acceptance	Â	3	5	30	1.83	No data		1.80	No data		2.63	4	13.3	4	23.3	60
		Â	9	18	162	1.84	6	1.7	1.70	»	»	1.90	14	8.64	20	12.3	61
		Σ	12	23	192	1.83	6	3.1	1.75	»	»	2.27	18	9.4	24	12.5	60.8
	Research tests	Â	3	3	18	1.81	1	5.6	1.61	»	»	2.08	5	27.8	6	33.3	33.3
		Â	4	4	30	1.92	5	16.7	1.74	2	6.7	1.78	5	16.7	12	40.0	33.3
		Σ	7	7	48	1.87	6	12.5	1.68	2	4.2	1.95	10	20.8	18	37.5	33.3
3	Acceptance	Â	1	2	15	2.30	No data		2.03	No data		2.40	2	13.3	2	13.3	50
		Â	1	2	15	1.80	»	»	2.50	»	»	2.60	2	13.3	2	13.3	0
		Σ	2	4	30	2.05	»	»	2.27	»	»	2.48	4	13.3	4	13.3	25
	Research tests	Â	1	1	9	1.80	»	»	2.10	»	»	2.20	No data				100
		Σ	1	1	9	1.80	»	»	2.10	»	»	2.20	»	»	»	»	100
5	Acceptance	Â	3	4	46	2.66	5	10.9	2.74	»	»	2.52	1	2.2	6	13.0	33.3
		Σ	3	4	46	2.66	5	10.9	2.74	»	»	2.52	1	2.2	6	13.0	33.3
	Research tests	Â	3	3	26	2.70	4	15.4	2.31	»	»	2.20	4	15.4	8	30.8	0
		Σ	3	3	26	2.70	4	15.4	2.31	»	»	2.20	4	15.4	8	30.8	0
1, 3, 5	Acceptance + research tests	Â	7	10	63	No data	1	1.6	No data	»	»	No data	11	17.5	12	19.0	50
		Â	17	29	288	»	20	6.9	»	2	0.7	»	26	9.0	48	16.6	58.3

Note. Here, O, S and G are, respectively, oxides, sulphides and globules.

**Table 7.** Contamination of steel ShKh15SG-V by non-metallic inclusions in evaluation by ASTM E-45 (acceptance inspection and research tests)

Research tests)										
Type of control	Variant of technology	Quantity, pcs		A_1	A_2	B_1	B_2	D_1	D_2	Output of efficient metal since first inspection, pcs / %
		Melting	Batch							
Billet										
Acceptance	A	6	6	$\frac{1.5-2.0}{1.75}$	$\frac{1.0-1.5}{1.33}$	$\frac{1.0-1.5}{1.42}$	$\frac{0.5-1.0}{0.92}$	1.0	1.0	6 / 100
	B	1	1	1.5	1.0	1.7	1.0	1.0	1.0	0 / 0
	Σ	7	7	$\frac{1.5-2.0}{1.63}$	$\frac{1.0-1.5}{1.17}$	$\frac{1.0-1.7}{1.56}$	$\frac{0.5-1.0}{0.96}$	1.0	1.0	6 / 85.7
Research tests	B	2	2	$\frac{1.1-1.2}{1.15}$	$\frac{0.9-1.0}{0.95}$	$\frac{1.4-1.5}{1.45}$	$\frac{1.0-1.2}{1.1}$	0.5	0.5	1 / 50
	Σ	2	2	$\frac{1.1-2.2}{1.15}$	$\frac{0.9-1.0}{0.95}$	$\frac{1.4-1.5}{1.45}$	$\frac{1.0-1.2}{1.1}$	0.5	0.5	1 / 50
Acceptance + research tests	A	6	6	$\frac{1.5-2.0}{1.75}$	$\frac{1.0-1.5}{1.33}$	$\frac{1.0-1.5}{1.42}$	$\frac{0.5-1.0}{0.92}$	1.0	1.0	6 / 100
	B	3	3	$\frac{1.1-1.2}{1.27}$	$\frac{0.9-1.0}{0.97}$	$\frac{1.4-1.7}{1.53}$	$\frac{1.0-1.2}{1.07}$	$\frac{0.5-1.0}{0.67}$	$\frac{0.5-1.0}{0.67}$	1 / 33.3
Ready metal products										
Acceptance	A	6	15	$\frac{1.5-2.0}{1.53}$	$\frac{1.0-1.5}{1.1}$	$\frac{1.5}{1.5}$	$\frac{0.5-1.0}{0.97}$	1.0	1.0	14 / 93
	B	2	2	$\frac{1.5}{1.5}$	$\frac{1.0}{1.0}$	$\frac{1.5}{1.5}$	$\frac{1.0}{1.0}$	1.0	1.0	2 / 100
	Σ	8	17	$\frac{1.5-2.0}{1.51}$	$\frac{1.0-1.5}{1.05}$	$\frac{1.5}{1.5}$	$\frac{0.5-1.0}{0.99}$	1.0	1.0	16 / 94.1
Research tests	A	3	3	$\frac{0.92-1.1}{1.01}$	$\frac{0.5-1.0}{0.67}$	$\frac{1.17-1.6}{1.42}$	$\frac{0.83-1.0}{0.89}$	$\frac{0.5-1.0}{0.72}$	$\frac{0.5-0.6}{0.53}$	2 / 66.7
	B	7	7	$\frac{0.9-1.5}{1.14}$	$\frac{0.5-1.0}{0.95}$	$\frac{1.1-2.1}{1.6}$	$\frac{0.7-1.43}{1.08}$	$\frac{0.5-1.0}{0.63}$	0.5	3 / 42.9
	Σ	10	10	$\frac{0.9-1.5}{1.1}$	$\frac{0.5-1.2}{0.87}$	$\frac{1.1-2.1}{1.55}$	$\frac{0.7-1.43}{1.02}$	$\frac{0.5-1.9}{0.94}$	$\frac{0.5-0.6}{0.51}$	5 / 50
Acceptance + research tests	A	9	18	$\frac{0.92-2.0}{1.44}$	$\frac{0.5-1.5}{1.03}$	$\frac{1.17-1.6}{1.49}$	$\frac{0.5-1.0}{0.96}$	$\frac{0.5-1.0}{0.95}$	$\frac{0.5-1.0}{0.92}$	16 / 88.9
	B	9	9	$\frac{0.9-1.5}{1.22}$	$\frac{0.5-1.2}{0.96}$	$\frac{1.1-2.1}{1.58}$	$\frac{0.7-1.43}{1.06}$	$\frac{0.5-1.0}{0.71}$	$\frac{0.5-1.0}{0.61}$	5 / 55.6

* Here, *A* — sulphides; *B* — oxides; *D* — globules (American standard).

metal was 19 %, and for metal of conventional melting — 24.8 %, that is stipulated by a smaller amount of rejections of experimental metal by oxides (1.6 %) against conventional metal (6.6 %).

Metal of variant *B* in comparison with metal of variant *A* is characterized by a decreased contamination with globular inclusions, whose share from total amount of rejections ($O + S + G$) is 54.2 % (by 1.7 times less than for metal of variant *A*) and increased contamination by oxides (6.9 against 1.6 % by variant *A*, see Table 6).

As to output of efficient metal, the metal of variant *B* is superior to metal of variant *A* (58.3 against 50 %) and better than the metal of conventional melting (55.1 %).

Thus, as a result of using imported 65 % ferrosilicium with a reduced content of calcium, the reduction of calcium in metal (by variant *B* of melting), as compared with conventional metal, from 9.2 (2006) (Figure 1) to 6.5 ppm (by 29 %), decrease in contamination by globular inclusions with growth in contamination by oxide inclusions was attained, while the output of efficient metal remained almost at the level of conventional metal.

Evaluation of metal by ASTM E-45, method A. As is seen from Table 7, the quality of metal of variant *A* is inferior to metal of variant *B* by sulphide inclusions (A_1 , A_2), however, it is superior by oxide inclusions (B_1 , B_2), that is coordinated with given results of investigations by GOST 801-73. Contami-

**Table 8.** Analysis of contamination of steel ShKh15SG, melted using ferrosilicium FS65 (DSTU 4127–2002), by non-metallic inclusions with evaluation by rates of GOST 801–78, — operating technology: November, December, 2006 and January, 2007

Group by GOST 801–78	RM; B	Quantity, pcs		Oxides			Sulphides			Globules			Output of efficient batches from the first inspection		O + S + G	
				Mean value	Rejections		Mean value	Rejections		Mean value	Rejections					
		Batch	Sample		pcs	%		pcs	%		pcs	%	pcs	%	pcs	%
1	B	76	1007	1.9	83	8.2	1.9	48	4.8	1.9	157	15.6	18	23.7	--	--
1	RM	109	1237	1.8	70	5.7	1.7	2	0.2	1.9	160	12.9	50	45.9	232	18.8
2	B	76	1007	1.9	34	3.4	1.9	9	0.9	1.9	70	7.0	46	60.5	--	--
2	RM	33	357	1.9	6	1.7	1.7	0	0	1.9	14	3.9	27	81.8	20	5.6
3	B	18	185	2.1	12	6.5	2.2	3	1.6	2.0	8	4.3	10	55.6	--	--
3	RM	39	329	2.2	18	5.5	2.0	4	1.2	2.1	22	6.7	26	66.7	44	13.4
4	B	18	185	2.1	0	0	2.2	0	0	2.0	0	0	18	100	--	--
4	RM	9	69	2.4	0	0	2.1	0	0	2.4	5	7.2	7	77.8	5	7.2
5	»	137	2012	2.5	51	2.5	2.7	106	5.3	2.6	146	7.3	81	59.1	303	15.6
1–5	»	327	4004	--	145	--	--	112	2.8	--	347	8.7	191	58.4	604	15.1
1, 3, 5	»	285	3578	--	139	3.9	--	112	3.1	--	328	9.2	157	55.1	579	16.2
1,3	»	148	1336	--	88	6.6	--	6	0.5	--	182	13.6	76	51.4	332	24.8

Note. Here, RM and B are, respectively, ready metal products and billet.

nation of metal by inclusions of type *D* is at the same level. Output of efficient metal by billet was 100 % for metal of variant *A* and 33.3 % for metal of variant *B*.

It was established during research tests that the metal of variant *B* of melting is inferior to metal of variant *A* by sulphides (A_1 , A_2) and oxides (B_1 , B_2), however, it is somewhat better by globules. As a whole, it is coordinated with results of evaluation of metal by GOST 801–78. Taking into account, that in evaluating by ASTM E-45 the characteristic by oxide is the most strict position, the output of the efficient metal by variant *A* was 88.9 %, and by variant *B* — 55.6 %.

CONCLUSIONS

1. Using the thermodynamic analysis, it was established that one of the sources of calcium enter to the metal, participating in formation of hard-to-deform (globular) alumocalcium inclusions is the processes of calcium reduction from slag melts by silicon and aluminium.

2. Using 65 % imported ferrosilicium with 0.08 % calcium content, 21 meltings of steel ShKh15SG-V were performed by two variants of ferrosilicium adding to metal: variant *A* — adding of ferrosilicium at metal tapping from furnace to ladle is 53 % of total amount of ferrosilicium and 47 % in IFL; variant *B* — adding of ferrosilicium at metal tapping from furnace into ladle — 95 % and in IFL — 5 %.

3. As a result of changing the scheme of ferrosilicium adding, a reduction in calcium content in ladle sample was attained, here, the calcium amount in ladle sample of metal of variant *A* was 11 ppm, variant *B* — 6.6 ppm.

4. Comparative investigation of contamination of experimental metal with evaluation by GOST 801–78 and ASTM E-45, method *A*, was made. It was established that, as compared with variant *A*, the metal of variant *B* at evaluation by GOST 801–78 is characterized by a decreased contamination by globular inclusions at simultaneous increase in content of oxide inclusions. As to the output of efficient metal in evaluation by GOST 801–78, some advantage of metal of variant *B* was outlined. In evaluation by ASTM E-45, method *A*, the metal of variant *A* of melting has the better characteristics by contamination as compared both with variant *B* and also metal of conventional melting.

5. It was established that application of imported ferrosilicium, in spite of low content of calcium (0.08 %), does not decrease the total contamination of steel by non-metallic inclusions both in its adding into ladle during metal tapping, and also in the process of treatment in IFL, but changes the structure of distribution of non-metallic inclusions into groups due to reduction of globular inclusions.

1. Gasik, M.I., Shulte, Yu.A., Gorobets, A.P. (1983) Physical-chemical processes of globular inclusion formation in bearing metal. *Izvestiya Vuzov. Chyorn. Metallurgiya*, **5**, 10–15.
2. Eliot, D.F., Glejser, M., Ramakrishna, V. (1969) *Thermochemistry of steelmaking processes*. Moscow: Metallurgiya.
3. Turkdogan, E.T. (1985) *Physical chemistry of high-temperature processes*. Moscow: Metallurgiya.
4. Morita, K., Sano, N. (2002) Activity measurement of slags equilibrated with molten silicon alloys. *Scandinavian J. of Metallurgy*, **31**, 178–183.
5. Grigoryan, V.A., Stomakhin, A.Ya., Ponomarenko, A.G. et al. (1989) *Physical-chemical calculations of electric steel-making processes*. Moscow: Metallurgiya.



CONCERNING THE ARTICLE OF G.A. TSYBULKIN «On issue of dependence of the carbon granular material electric properties upon pressure and size of fractions»

V.I. LAKOMSKY

E.O. Paton Electric Welding Institute, NASU, Kiev, Ukraine

In the second issue of our journal for 2007, the article of G.A. Tsybulkin «On issue of dependence of the carbon granular material electric properties upon pressure and size of fractions» was published. Analysis of this article made me to express my opinion on the information described in it.

G.A. Tsybulkin presented a purely mathematical exercise to the detriment of a physical principle of electric contact phenomena of solid carbon materials.

The history of development of methods of production of thermoanthracite turned out so that, in spite of 25-year service of furnaces for production of thermoanthracite, nobody studied experimentally a specific electric resistance (SER) of a compact thermoanthracite, and also the relation between SER of the compact and fraction composition of a bulk mixture.

The first work, directed to the search of analytic relation between SER of a lumpy electrically-conductive material and its fractional composition, i.e. a real subject of article of G.A. Tsybulkin, was published by me in 2003 [1]. Utilitarian aim of this work was to obtain a calculation method of determination of a numerical value of SER of a thermoanthracitic charge (+6 – –25 mm) of electric calcinators of Dnepropetrovsk electrode plant. The desire to spread the obtained method of calculation for fine powder-like mixtures has led to the fact that G.A. Tsybulkin and me have fulfilled and published the second work [2].

Both in the first and second articles, the considered physical phenomenon of electric contacting of solid carbon materials was based on the works of R. Holm, the known German scientist [3]. This scientist by his works in the field of electric contacts deserved recognition and prestige in electrophysicists of the whole world. All his life he worked at the research center of the well-known electric company AEG. His research works, made in laboratories of the center, are characterized by a very profound experiment and deepness of a theoretical analysis of phenomena.

By study the electric contacts of solid bodies, R. Holm made a great innovative contribution to the physics of solid body contacts. In his works, he used the results of study of a microrelief surface of me-

chanically treated solid bodies by various scientists, as well as a classic work of D. Maxwell, in which he was the first who showed that there is a contraction resistance in nature, except the widely-known volume electric resistance of a conductor material. It occurred, that if to make contractions in the electric conductor of a certain cross section, then the abrupt narrowing or widening of electric conductor will cause contraction or widening of electric power lines of current in the conductor body, for which a definite energy of electromagnetic field, equivalent to the electric resistance of contraction, is consumed. Coming from this, R. Holm introduced a new concept of an elementary α -spot on a contact surface of solid body into the physics of electric contacts [3].

The mentioned α -spot by R. Holm represents a spot of a real physical contact of opposite projections on micro-relieves of two contacting surfaces of the solid body. Size of the elementary α -spot is very small, and shape of a real contact surface, as R. Holm assumed, is round (later it was proved that the shape of a contact spot, in particular on brittle materials, is really round [4]). Electric current, passing from one fragment of a solid body into another through this spot of contact, in which the density of current reaches $1 \cdot 10^6$ A / cm², overcomes the contraction resistance R_c . For a contact pair this resistance is determined very simply: ρ_0 / a , where ρ_0 is the specific electric resistance of contact material; a is the radius of elementary contact spot.

Electric contact conception of R. Holm was recognized by scientists of the whole world not only in the field of electric engineering, but also in other branches of knowledge, studying the phenomena of energy transfer through the solid body interface [4].

According to R. Holm, not one but n amount of α -spots are arranged on visible (apparent) area of a solid-body contact, therefore, the contraction resistance of the contact as a whole will be lower by n times. As in a real practice the numerical value n is not known, then the given value a , which is easy determined from a simple experiment, is used in the engineering calculations. For example, the specialists of Dnepropetrovsk electrode plant and me had to determine the numerical value of a radius of a given α -spot on the thermoanthracite mixture of a fraction



composition +4 – –10 mm, using industrial equipment for electric preheating of thermoanthracite charge before its mixing with a coat-tar pitch. At shop temperature the arithmetic mean value a was $(8.5 \pm \pm 0.97) \mu\text{m}$ for the conditions of so-called thin layer of a grainy thermoanthracite [5].

R. Holm clearly assumed by proposing his conception of a -spot that two elements in a contact pair are obligatory pressed one to another at a definite force P . This is natural, and, therefore, he did not dwell on this specific feature of the contact pair. Evidently, in this connection he did not give the analytical dependence of a on P in his monograph [3].

As to the effect of pressure on size of elementary a -spot on the surface of such brittle materials as, for example, anthracite and its derivative, i.e. thermoanthracite, then it is necessary to distinguish low and high pressures, that was considered by G.A. Tsybulkin in his article. Low pressures include pressures, which do not lead to fracture (crushing) of grains of the material studied. In this case the fraction composition of a weighed quantity of the test material remains unchanged in the process of testing. By my observations, the boundary between low and high pressure is 0.35 atm.

As we know, low pressures have not been yet studied by anybody. In monograph [6], on which G.A. Tsybulkin referred, only high pressures were studied, at which SER of powdered carbon material reached the value of SER monolith, that is several thousands of atmospheres.

In work [7], I had to make the first steps in study of effect of low pressures on electric resistance of a crushed anthracite. In this work I presented three assumed structures of elementary contact spots on thermoanthracite. Variety of structure of a -spots on the thermoanthracite surface is explained by a structure anisotropy of thermoanthracite the same as of the anthracite [8]. Due to the above anisotropy we had to make hundred measurements in study of dependence of SER of a crushed anthracite (+4 – –6 mm) on low pressures, until we could distinguish three different types of dependencies. In all studied dependencies $\rho_b = f(P)$, SER of a crushed thermoanthracite ρ_b is decreased with growth in pressure, it is decreased very intensively in the first moment of load applying. It seemed not possible to suggest a common dependence for all three variants of structures of a -spots, because of serious differences between the dependencies studied.

It should be outlined that if the SER of bulk medium of a solid body is considered, then there is no need to note specially the contact pressure. According to mechanics of bulk media [9, 10], each layer of a bulk mass experiences its contact pressure depending on the depth of location. Coming from this, when measuring SER of a crushed material in case of its free location, the concept of «thin layer» is accepted. «Thin layer» is accepted as such a layer of bulk crushed material whose effect on results of SER measurements can be neglected. In case of work [5] the thickness of the above-mentioned layer was 250 mm.

The given brief description of conception of R. Holm did not satisfy G.A. Tsybulkin by some reasons and he suggested his conception about the nature of contacts of two solid bodies. Gennady Aleksandrovich Tsybulkin considers a contact spot between fragments of solid body as a result of elastic or plastic deformation of the given solid body. And this is referred by the author to the thermoanthracite, as is seen from the article text. There is a question, about what elastic deformation of the anthracite are they speaking, if its hardness by Mohs is 7–8 units [8] and even higher. This material can «cut glass». Further on, applying equation (1) obtained in work [2] using conception of R. Holm, he criticizes R. Holm himself and obtains on the basis of mathematical theory of elasticity a dependence of radius of a -spot on external pressure P to a contacting pair of a radius of grain of carbon material r and elastic properties of material of contact E and ν :

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

Here, it is worthy to note that inquisitive reader will not find the numerical values E and ν for such brittle and anisotropic material as thermoanthracite in any handbooks on carbon materials. Only numerical values E and ν are known for single crystal of graphite and isotropic fine-grained graphitized materials [11, 12], moreover, it is indicated almost in each handbook that graphite is quite elastic body and given values E and ν are obtained mostly by measurement of rate of sound spreading in the material.

It follows from the equation that the larger the size of grain of contacting carbon material, the larger size of elementary a -spot. May be it is really so, but if not to indicate the limits of application of this relation, then it will lead to absurdity. In accordance with GOST 4794–97, maximum size of bulk charge of the thermoanthracite can reach 120 mm. So, what size can be reached by a contact spot, if to use the formula suggested by G.A. Tsybulkin? A millimeter order? Then, what numerical value of contraction resistance will be? Too negligible? But this contradicts the reality, as the electric resistance of a crushed material, in particular, of a coarse crushing, will be always higher than the resistance of a compact.

The suggested dependence, from my point of view, is not subjected to a physical interpretation. According to R. Holm everything is going on simpler. Real physical contact is commenced at the moment, when the highest projections on microrelief contact surfaces are contacted in contact jaws bringing together. If we are speaking about copper surfaces, then during further bringing together of contacting surfaces, so first the process of elastic and then elastic-plastic and, finally, plastic deformation is proceeding. Bringing together of contacting jaws will continue until the force of clamping of contact jaws is not be equal to the force of reaction of the metal being deformed.



This process is described in detail by me in a brief monograph [13].

The most interesting in article of G.A. Tsybulkin is the fact that he received an excellent coincidence of results of experiments, carried out on thermoanthracite, with a theoretical curve (see Figure in article of G.A. Tsybulkin) by taking a known-well unacceptable initial condition for brittle electrically-conductive material in deriving the analytic relationship. This excellent coincidence of theoretical and experimental data are rarely observed in laboratory investigations. In the given case this fact puts us on sad thoughts about giving desirable instead of real.

Thus, I would like to attract attention of the reader once again to purely mathematical approach of G.A. Tsybulkin to the solution of a rather complicated problem, to his ignoring the physical analysis of obtained mathematical conclusions. In this connection, the statement of Gibbs, the well-known American physicist-theoretic, should be remembered. Some more than 100 years ago one of the founders of chemical thermodynamics said that «mathematics is the language». And not more. Using mathematics the scientists transfer each other the complex phenomena of nature, studied by them.

1. Lakomsky, V.I. (2003) Mathematical model of calculation of specific electrical resistance of grained thermoanthracite depending on its fractional composition. *Advances in Electrometallurgy*, **3**, 44–47.
2. Lakomsky V.I., Tsybulkin, G.A. (2005) Analytical model of specific resistance of electroconductive bulk material. *Ibid.*, **1**, 40–42.
3. Holm, R. (1961) *Electric contacts*. Moscow: Inostr. Literatura.
4. Shlykov, Yu.P., Ganin, E.A., Tsarevsky, S.N. (1977) *Contact thermal resistance*. Moscow: Energiya.
5. Bykovets, V.V., Lakomsky, V.I., Kirilenko, V.P. (2004) Specific electric resistance of lumpy thermoanthracite in thin layer. *Advances in Electrometallurgy*, **1**, 44–45.
6. Agroskin, A.A. (1959) *Thermal and electric properties of coals*. Moscow: Metallurgizdat.
7. Lakomsky, V.I. (2006) Dependence of specific electric resistance of lumpy thermoanthracite on pressure. *Dopovidi Nats. Akad. Nauk Ukrainy*, **4**, 96–103.
8. Fialkov, A.S. (1977) *Carbon, interlayer compounds and composites on its base*. Moscow: Aspect Press.
9. Tertsagi, K. (1961) *Theory of mechanics of grounds*. Moscow: Gosstrojizdat.
10. Sokolovsky, V.V. (1965) *Mechanics of granulated solids*. Moscow: Gosstrojizdat.
11. Ubellode, A.R., Liuis, F.A. (1965) *Graphite and its crystalline compounds*. Moscow: Mir.
12. Ostrovsky, V.S., Vrgiliev, Yu.S., Kostikov, V.I. et al. (1966) *Artificial graphite*. Moscow: Metallurgiya.
13. Lakomsky, V.I. (2007) *Electric and electric contact properties of thermoanthracite*. Kiev: Akadempriodika (in the press).

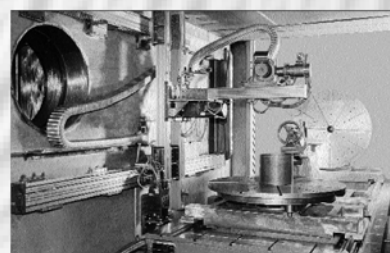
RANGE OF KL-109, KL-110 AND KL-111 UNIVERSAL MACHINES FOR EBW OF LARGE AND HEAVY WORKPIECES

- PC and programmable controllers are used.
- Electron beam parameters analysis and «black box» type self-diagnostics of machine by PC.
- Real-time seam tracking and monitoring of EBW process by RASTR system on the basis of the secondary electron emission.
- Gun power source with electron tube flashless system.

Mobile type 15, 30 or 60 kW electron beam gun at accelerating voltage of 60 kV.

Design

The work chamber has two sliding doors. The workpiece table is moved out of the work chamber onto the runout platform of EBW process. The table accommodates rotators with horizontal and vertical axis, and also back centre. The electron gun 3-axis-manipulator has the travelling distance in X-direction up to 3000 mm, in Y-direction up to 730 mm and in Z-direction up to 1500 mm. Precision of the guidance and drive system equals that of precision machine tools operation with tolerances in the hundredth-of-a-millimeter range.



Contacts: Prof. Nazarenko O.K.
E-mail: nazarenko@technobeam.com.ua



REPAIR BRAZING OF PARTS OF THE TURBINE HOT PATH

Recently, the repair brazing of nickel heat-resistant alloys (HRA) is used instead of the process of argon arc welding in restoration of components of a hot part of aircraft and industrial (power) gas-turbine plants. First of all, it concerns the sub-assemblies of a guiding unit of turbines and combustion chambers, having a system of perforated holes providing the film cooling of surface of parts being repaired, flaps of an adjustable nozzle, etc. In welding, the irreversible geometric changes of separate components and violation of seats are occurred due to high temperature of metal. Welding causes «clogging» of perforated holes, and it is necessary to renew them using in most cases the piercing by a laser beam.

In brazing, the hazardous residual thermal stresses and undesirable changes of a geometric shape of parts being restored are not occurred. Contact-reaction brazing of nickel HRA is equivalent by its technological capabilities to the argon arc welding and provides the required physical-mechanical properties of the joints.

The application of the repair brazing is based on the effect of an activated diffusion interaction between particles of a composite brazing alloy and surfaces joined, being a result of a contact fusion of surface of particles and part wall by a low-melting component of the brazing alloy, and also by a post heat treatment of the brazed joint at temperature, lower by 80–120 °C than maximum (approx. 1230 °C) temperature of the isothermal brazing.

Temperature and duration of processes of brazing and heat treatment should be coordinated with conditions of scheduled technological operations (annealing) of definite products of appropriate alloys.

Strength of brazed joints of different heat-resistant alloys ($T_{\text{test}} = 20\text{ }^{\circ}\text{C}$)

Type of alloy	Mean value σ_t , MPa		$\sigma_t^{\text{BJ}} / \sigma_t^{\text{BM}}$
	Base metal	BJ*	
ChS70VI	810	732 / 778	0.90 / 0.96
VJL12U	830	721 / 873	0.86 / 1.05
JS26VI	730	692 / 718	0.95 / 0.99
JS26NK	870	766 / 835	0.88 / 0.96

* Here, the numerator shows σ_t of brazed joints produced using boron-containing brazing alloy, the denominator shows σ_t of brazing alloy containing additionally 20 wt.% NS12.

With appearance in service of new HRA, it is necessary to develop new repair materials: brazing alloys and proper fillers. Material of the brazed weld should be high-strength, capable to withstand shock loads, its microstructure should correspond to a large

extent to the structure of metal of product being repaired. The latter is guaranteed by application in repair of components, close by chemical composition, and vacuum heat treatment of products before brazing (at temperature of process of isothermal brazing).

It is rational to use brazing alloys with their maximum similar chemical composition to that of the part being repaired. This is attained by adding of powders of alloys to be brazed or more high-alloy analogs to used brazing alloys as a filling material.

The repair brazing of parts of nickel HRA is made by complex brazing alloys, containing materials, as a filler, providing a sufficient level of a corrosion resistance together with a satisfactory strength and fracture toughness. These characteristics are defined by chemical and phase compositions of metal of the brazed joint and confined zone of a thermal diffusion interaction of weld with a base metal

The Research Center «Pratt & Whitney Paton» has developed the technological approaches of repair brazing (restoration) of parts of a hot path of turbines made from cast nickel alloys ChS70VI, VJL12U, EP-648, ChS104, JS6U, JS26VI and NK, etc. after service. They are based on application of modified silicon- and boron-containing brazing alloys with powder fillers of nickel HRA. The improvement of wettability of brazed surfaces by a melt, reliable penetration and filling of gaps and capillaries in damaged parts by a melt were attained.

Peculiarities, typical of brazed joints made with use of silicon- and boron-containing complex brazing alloys, consist in a limited reaction behavior of the above-mentioned alloys with respect to the base metal, minimum porosity, absence of shrinkage defects in weld, dispersion of embrittling carbide-boride phases.

Statistic processing of results of tensile tests of brazed joints at room temperature showed a noticeable increase in a level of strength, elongation and, most important, the stability and repetition of test results. More than 65 % of samples, brazed by complex brazing alloys, have an ultimate strength, corresponding to that of the base metal.

The Table shows the functional data for brazed joints of some cast nickel superalloys, the parts of which have passed successfully the restoration by brazing and post two-step heat treatment in vacuum. Soundness of brazed joints of different alloys at room temperature is 0.86–1.05, while at $T = 800\text{--}1000\text{ }^{\circ}\text{C}$ it reaches 0.80–0.97.

The developed systems of complex silicon- and boron-containing brazing alloys are universal for brazing of parts made from nickel superalloys. Depending on the alloy type, only temperature and duration of isothermal brazing of products are changed.

INDEX OF ARTICLES FOR AEM'2007, Nos. 1–4

ELECTROMETALLURGY OF STEEL AND FERROALLOYS

Comparative experimental-industrial investigations of effect of 65 % ferrosilicium with different content of calcium on contamination of steel ShKh15SG-V with globular alumocalcium inclusions (Panchenko A.I., Salnikov A.S., Logozinsky I.N., Korol L.N., Kazakov S.S., Gasik M.I. and Gorobetz A.P.)

Experience of producing especially low-carbon steel for plastic wire rod (Saviuk A.N., Derevyanchenko I.V., Kucherenko O.L., Projdak Yu.S., Stovpchenko A.P., Kamkina L.V. and Grishchenko Yu.N.)

Mastering of production technology and investigation of quality of especially low-carbon rolled wire (Saviuk A.N., Derevyanchenko I.V., Kucherenko O.L., Stovpchenko A.P., Projdak Yu.S., Kamkina L.V. and Grishchenko Yu.N.)

State and vectors of development of electric steel production in Ukraine (Efimenko G.G. and Postizhenko V.K.)

ELECTRON BEAM PROCESSES

Commercial electron beam installation UE-5812 (Trigub N.P., Zhuk G.V., Kornejchuk V.D., Ishchuk Yu.T., Severin A.Yu. and Davydov S.V.)

Effect of substrate temperature on micro- and substructure of copper condensates deposited from a vapor phase (Ustinov A.I., Fesyun E.V., Melnichenko T.V. and Romanenko S.M.)

Electron beam method of graphite evaporation and production of condensates free from tungsten impurities (Kurapov Yu.A. and Movchan B.A.)

Hard erosion-resistant graded coatings deposited in vacuum (Yakovchuk K.Yu., Skryabinsky V.V., Marinsky A.G. and Didikin G.G.)

Investigation of a number of regularities of electron beam evaporation and condensation of carbon (Movchan B.A., Kurapov Yu.A. and Krushinskaya L.A.)

Investigation of distributed characteristics of electron beam formed by ring cathode in electron beam zone melting (Demchenko V.F., Asnis E.A., Lesnoj A.B., Zabolotin S.P. and Shegelsky N.E.)

Modeling of evaporation processes of alloying elements from Ti–6Al–4V alloy in EBCHM and portioned discharge (Akhonin S.V., Milenin A.S. and Samofalov A.V.)

Morphology of carbides and microstructure of steel R6M5 of electron beam remelting (Shpak P.A., Grechanyuk N.I., Osokin V.A. and Artemchuk A.A.)

Reconstruction of electron beam installation TICO-15M (Kostenko V.I., Pap P.A., Kalinyuk A.N., Kovalchuk D.V., Kondraty N.P. and Chernyavsky V.B.)

ELECTROSLAG TECHNOLOGY

Certain aspects of deposited layer formation in portioned electroslag cladding with application of liquid additive metal (Tsykulenko K.A.)

Electroslag casting with melting together. Review (Zhadkevich M.L., Shevtsov V.L. and Puzrin L.G.)

New technological process of producing super-large steel ingots by ESC LM method (Paton B.E., Medovar L.B., Saenko V.Ya., Tsykulenko A.K., Fedorovsky B.B. and Us V.I.)

On new approach to construction of the ESR automatic control system (Troyansky A.A., Ryabtsev A.D., Petrenko V.L., Medovar L.B., Dymnich A.Kh., Mastepan V.Yu. and Ratiev S.N.)

Production of increased strength titanium by its doping with oxygen in process of chamber electroslag remelting (Ryabtsev A.D., Davydov S.I., Troyansky A.A., Shvartsman L.Ya., Ryabtsева O.A., Pashinsky V.V. and Feofanov K.L.)

Production of vanadium-containing hardeners with application of electroslag technology (Biktagirov F.K., Khudik B.I., Zhadkevich M.L., Prokhorenko K.K., Shapovalov V.A., Korostin A.D. and Muratov V.M.)

Progress of electroslag technologies and updating of designs of ESR moulds (Review) (Tsykulenko K.A.)

Towards the problem of intensification of electromagnetic action in magnetically-controlled electroslag melting of titanium alloys (Kompan Ya.Yu., Nazarchuk A.T. and Protokovilov I.V.)

ENERGY AND RESOURCES SAVING

Alternative technologies of remelting of industrial wastes of titanium and alloys thereof (Shejko I.V., Shapovalov V.A. and Konstantinov V.S.)

Increase of utilization factor of refractory alloy waste by electroslag methods (Maksyuta I.I., Kvasnitskaya Yu.G., Simanovsky V.M. and Myalnitsa G.F.)

On issue of dependence of the carbon granular material electric properties upon pressure and size of fractions (Tsybulkin G.A.)

1 GENERAL PROBLEMS OF METALLURGY

Analysis of the state-of-the-art and problems of development of continuous casting of steel in metallurgy of Ukraine (Postizhenko V.K. and Efimenko G.G.)

2 Diffusion coatings on the base of titanium, vanadium and chromium carbides on steel U8A (Khizhnyak V.G., Pomarin Yu.M., Kurilo N.A. and Medova I.Yu.)

1 Investigation of diffusion processes in multilayer composite thermal coatings (Borisov Yu.S., Adeeva L.I., Kaplina G.S., Tunik A.Yu., Gordan G.N., Demianov I.A. and Revo S.L.)

1 Mechanical properties and structure of brazed joints of casting nickel alloy JS26VI. Part 1 (Malashenko I.S., Kurenkova V.V., Onoprienko E.V., Trokhimchenko V.V., Belyavin A.F. and Chervyakova L.V.)

4 Mechanical properties and structure of brazed joints of JS26VI cast nickel alloy. Part 2 (Kurenkova V.V., Onoprienko E.V., Malashenko I.S., Grabin V.V., Belyavin A.F. and Chervyakova L.V.)

3 Physical-chemical processes in composite powder particles under conditions of plasma spraying (Borisov Yu.S., Borisova A.L., Tunik A.Yu. and Murashov A.P.)

2 Structure and strength properties of brazed joints of cast nickel alloy JS26NK. Part 1 (Malashenko I.S., Kurenkova V.V., Belyavin A.F., Trokhimchenko V.V., Chervyakova L.V. and Lebedevich Ya.B.)

1 Titanium. Problems of production. Prospects. Analytical Review. Part 1 (Tsykulenko K.A.)

3 Titanium. Problems of production. Prospects. Analytical Review. Part 2 (Tsykulenko K.A.)

2 Titanium. Problems of production. Prospects. Analytical Review. Part 3 (Tsykulenko K.A.)

INFORMATION

1 Congratulate with jubilee

Developed at PWI

3 Prof. M.L. Zhadkevich is 70

3 Repair brazing of parts of the turbine hot path

PLASMA-ARC TECHNOLOGY

Investigation of composition of burnt gas in plasma liquid-phase reduction of iron from ore raw materials by gaseous reducers (Zhadkevich M.L., Shapovalov V.A., Melnik G.A., Zhirov D.M., Zhdanovsky A.A., Tsykulenko K.A. and Vislobokov O.M.)

2 Measurement of temperature of drum-cooler in the process of super-rapid hardening of metal melted in plasma furnace (Shapovalov V.A., Nikitenko Yu.A., Burnashev V.R. and Torkhov G.F.)

1 Power parameters of process of plasma liquid-phase reduction of iron (Zhadkevich M.L., Shapovalov V.A., Zhirov D.M., Melnik G.A., Prikhodko M.S. and Zhdanovsky A.A.)

REVIEWS AND SCIENTIFIC DISCUSSIONS

Concerning the article of G.A. Tsybulkin «On issue of dependence of the carbon granular material electric properties upon pressure and size of fractions» (Lakomsky V.I.)

VACUUM-INDUCTION MELTING

3 Distribution of oxygen in single crystals of silicon and its influence on life time of non-equilibrium charge carriers (Chervony I.F., Shvets E.Ya., Volyar R.N. and Golev A.S.)

4 Index of articles for AEM'2007, Nos. 1–4

List of authors

LIST OF AUTHORS

Adeeva L.I. No.3
Akhonin S.V. No.2
Artemchuk A.A. No.1
Asnis E.A. No.3

Belyavin A.F. No.1, 2, 4
Biktagirov F.K. No.3
Borisov Yu.S. No.2, 3
Borisova A.L. No.2
Burnashev V.R. No.4

Chernyavsky V.B. No.3
Chervony I.F. No.1
Chervyakova L.V. No.1, 2, 4

Davydov S.I. No.3
Davydov S.V. No.1
Demchenko V.F. No.3
Demianov I.A. No.3
Derevyanchenko I.V. No.1, 2
Didikin G.G. No.2
Dymnich A.Kh. No.2

Efimenko G.G. No.1, 4

Fedorovsky B.B. No.1
Feofanov K.L. No.3
Fesyun E.V. No.4

Gasik M.I. No.4
Golev A.S. No.1
Gordan G.N. No.3
Gorobetz A.P. No.4
Grabin V.V. No.2
Grechanyuk N.I. No.1
Grishchenko Yu.N. No.1, 2

Ishchuk Yu.T. No.1

Kalinyuk A.N. No.3
Kamkina L.V. No.1, 2
Kaplina G.S. No.3
Kazakov S.S. No.4
Khizhnyak V.G. No.4
Khudik B.I. No.3
Kompan Ya.Yu. No.4
Kondraty N.P. No.3
Konstantinov V.S. No.3

Kornejchuk V.D. No.1
Korol L.N. No.4
Korostin A.D. No.3
Kostenko V.I. No.3
Kovalchuk D.V. No.3
Krushinskaya L.A. No.1
Kucherenko O.L. No.1, 2
Kurapov Yu.A. No.1, 3
Kurenkova V.V. No.1, 2, 4
Kurilo N.A. No.4
Kvasnitskaya Yu.G. No.1

Lakomsky V.I. No.4
Lebedevich Ya.B. No.4
Lesnoj A.B. No.3
Logozinsky I.N. No.4

Maksyuta I.I. No.1
Malashenko I.S. No.1, 2, 4
Marinsky A.G. No.2
Mastepan V.Yu. No.2
Medova I.Yu. No.4
Medovar L.B. No.1, 2
Melnichenko T.V. No.4
Melnik G.A. No.1, 3
Milenin A.S. No.2
Movchan B.A. No.1, 3
Murashov A.P. No.2
Muratov V.M. No.3
Myalnitsa G.F. No.1

Nazarchuk A.T. No.4
Nikitenko Yu.A. No.4

Onoprienko E.V. No.1, 2
Osokin V.A. No.1

Panchenko A.I. No.4
Pap P.A. No.3
Pashinsky V.V. No.3
Paton B.E. No.1
Petrenko V.L. No.2
Pomarin Yu.M. No.4
Postizhenko V.K. No.1, 4
Prikhodko M.S. No.3
Projdak Yu.S. No.1, 2
Prokhorenko K.K. No.3
Protokovilov I.V. No.4

Puzrin L.G. No.3

Ratiev S.N. No.2
Revo S.L. No.3
Romanenko S.M. No.4
Ryabtsev A.D. No.2, 3
Ryabtseva O.A. No.3

Saenko V.Ya. No.1
Salnikov A.S. No.4
Samofalov A.V. No.2
Saviuk A.N. No.1, 2
Severin A.Yu. No.1
Shapovalov V.A. No.1, 3(3), 4
Shegelsky N.E. No.3
Shejko I.V. No.3
Shevtsov V.L. No.3
Shpak P.A. No.1
Shvartsman L.Ya. No.3
Shvets E.Ya. No.1
Simanovsky V.M. No.1
Skryabinsky V.V. No.2
Stovpchenko A.P. No.1, 2

Torkhov G.F. No.4
Trigub N.P. No.1
Trokhimchenko V.V. No.1, 4
Troyansky A.A. No.2, 3
Tsybulkin G.A. No.2
Tsykulenko A.K. No.1
Tsykulenko K.A. No.1(2), 2(2), 3, 4
Tunik A.Yu. No.2, 3

Us V.I. No.1
Ustinov A.I. No.4

Vislobokov O.M. No.1
Volyar R.N. No.1

Yakovchuk K.Yu. No.2

Zabolotin S.P. No.3
Zhadkevich M.L. No.1, 3(3)
Zhdanovsky A.A. No.1, 3
Zhiron D.M. No.1, 3
Zhuk G.V. No.1