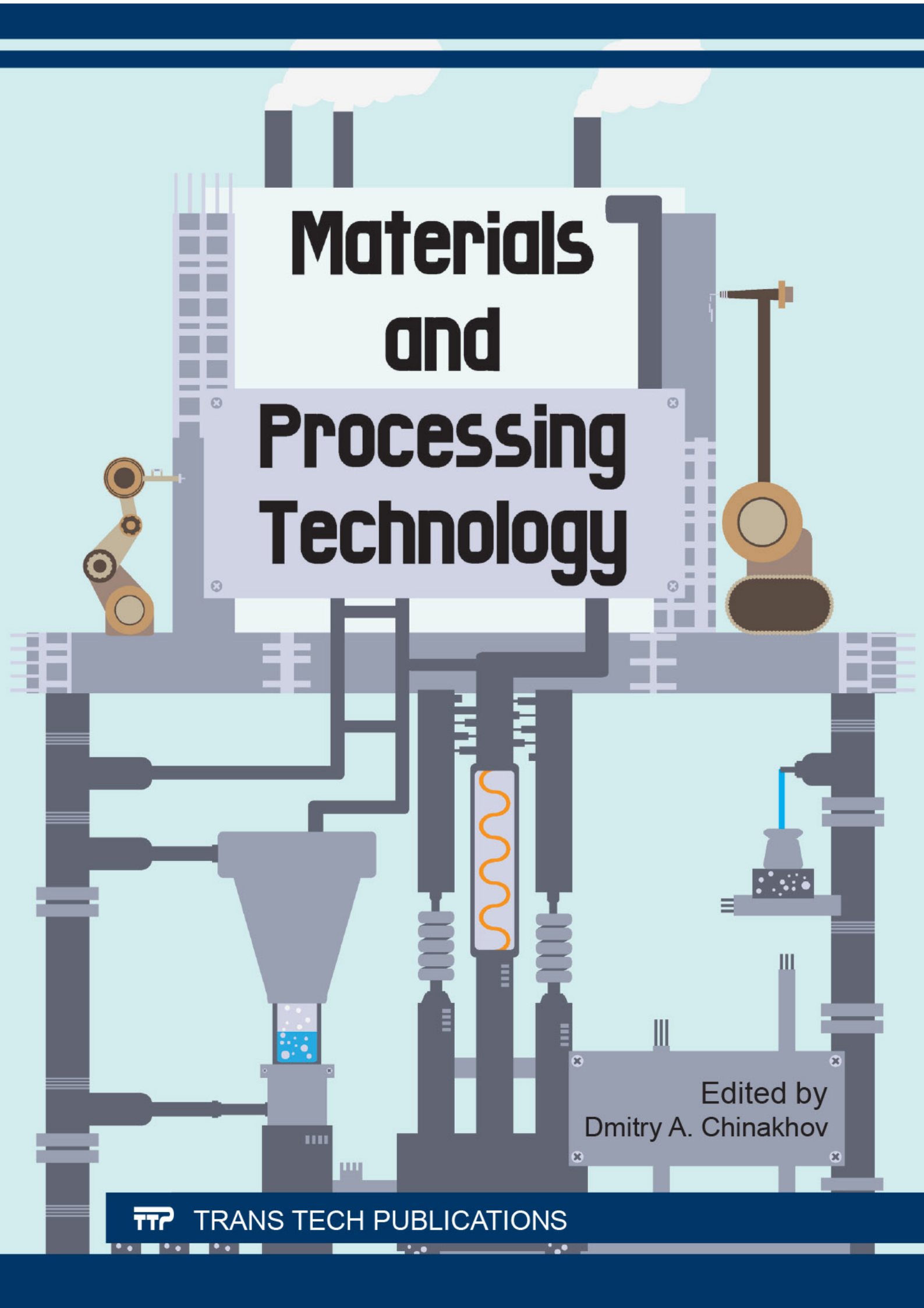


The background is a stylized industrial scene. At the top, three smokestacks emit white smoke against a light blue sky. Below them, a complex network of grey pipes and machinery is depicted. On the left, a robotic arm with orange joints is visible. In the center, a large white rectangular area contains the title text. To the right, a crane-like structure with a brown arm and a circular base is shown. The overall aesthetic is clean and modern, using a limited color palette of greys, blues, and oranges.

Materials and Processing Technology

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Preface

Intensive development of engineering contributes to the emergence of new materials and technologies for their processing. This special issue contains research papers on modern technologies for obtaining and processing materials, technologies for obtaining welded joints, additive technologies.

The book is intended for a wide range of specialists engaged in the development and production of heavy-duty metal structures, as well as students, undergraduates, graduate and postgraduate students of technical colleges and universities.

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Vice-Chancellor, Yurga Institute of Technology
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Abrasive Wear Resistance Comparative Analysis of the Metal Surfaced by Flux Cored Wires Systems Fe-C-Si-Mn-Ni-Mo-W-V and Fe-C-Si-Mn-Cr-Ni-Mo-V

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Keywords: flux cored wire, abrasive-shock wear, surfacing, recovering, wear resistance, hardness, nonmetallic inclusions, structure.

Abstract. In the article is shown the comparative analysis between structures of surfaced by the flux coded wire metal systems Fe-C-Si-Mn-Cr-Ni-Mo-V and Fe-C-Si-Mn-Ni-Mo-W-V. These powder wires are supposed to be used in recovering details and equipment components and machines, that works in conditions of intensive abrasive – shock wear. Manufacturing and surfacing of flux cored wires samples were made in laboratory conditions. Defined chemical composition of the surfaced metal. Deposited metal samples hardness and wear resistance were researched. In the course of deposited meta surface metallographic analysis were made following metallographic researches: defined nature and level of nonmetallic oxides impurity, type and morphology of the microstructure, grain size of surfaced samples. Estimation of the chemical composition components influence on the hardness and wear resistance were obtained.

Introduction

Developing new materials and innovation recovering technologies, that extremely increase items wear resistance property, is the current problem in engineering. For these purpose development and manufacturing of specialized methods and materials of surfacing is in progress [1-16]. Most popular for surfacing detail which works in abrasive wear conditions is flux cored wire for surfacing of low-carbon and low-allow type, austenitic and high-manganese type C, and rapid steels type F per IIC classification [17]. Also, surfacing carbide allows type P have a widespread use, that constitute composite materials, and comprise reinforcing tungsten carbide elements and matrix. They are notable for high wear resistance in abrasive-shock wear conditions [17]. The characteristic feature in the wearout prosses for such allows is stepwise separate elements wearout of composition. In that case appears what is known as shading affect, when most wear resistance reinforcing elements took all major pressure from destructive force on itself, protecting allow matrix from wearout. Then, while matrix wear resistance is in balance its components working capacity could be defined by its chemical composition, concentration, wear resistance and durability of the reinforce elements [1]. However, matrix wear resistance could be the key factor of detail working capacity in abrasive wear conditions.

Materials and Technology

In these work is shown the hardness and wear resistance research results of the metal which is surfaced whit the use of flux coder wire systems Fe-C-Si-Mn-Cr-Ni-Mo-V and Fe-C-Si-Mn-Ni-Mo-W-V.

Flux cored wire manufacturing was conducted on laboratory machine. Diameter of powder wire is 5mm, wire shell is made from metal ribbon of St3 grade. As a filler were used corresponding powder materials: iron powder brand PJV 1 by GOST 9849-86, ferrosilicon powder brand FS 75 by GOST 1415-93, powder of high carbon ferrochrome brand FH900A by GOST 4757-91, powder of carbon ferromanganese FMn 78(A) by GOST 4755-91, nickel powder PNK-1L5 by GOST 9722-97, ferromolybdenum powder brand FMo60 by GOST 4759-91, tungsten powder brand PVP-1 by

TI 48-19-71-78, ferrovanadium powder brand FV50U 0.6 by GOST 27130-94, cobalt powder brand PK-1U2 by GOST 9721-79. As a carbon-contained component was used carbon-fluorine contain material (CFCM) that was previously used in works [18-21] with the following chemical composition, wt.%: $\text{Al}_2\text{O}_3 = 21-46.23$; $\text{F} = 18-27$; $\text{Na}_2\text{O} = 8-15$; $\text{K}_2\text{O} = 0.4-6$; $\text{CaO} = 0.7-2.3$; $\text{Si}_2\text{O} = 0.5-2.48$; $\text{Fe}_2\text{O}_3 = 2.1-3.27$; $\text{C}_{\text{obsh}} = 12.5-30.2$; $\text{MnO} = 0.07-0.9$; $\text{MgO} = 0.06-0.9$; $\text{S} = 0.09-0.19$; $\text{P} = 0.1-0.18$.

Surfacing under the AN-26S flux was conducted in 6 layers on samples of the 09G2S grade steel with its 16mm thickness. Prosses was conducted with the use of automatic welding machine ASAW-1250 under following welding conditions: $I_w=500$ A; $U_a=30$ V; $V_w=9$ meter per hour.

Hardness of the researched samples was measured with the use of hardness meter MET-DU. Hardness measurements were performed 5 times on the surface of each sample. In table 1 is shown average value of surfaced metal hardness measurements. Wear resistance tests were conducted no the laboratory machine 2070 SMT – 1. Wear tests were conducted as per flowchart rotating disk – sample. The disk has tungsten allow hard surface. Every test took 6 hours with rotation frequency 20 revolutions per minute. Before and after wear tests the samples were weighted to define the difference between initial and final weight, and the disk revolution amount of each test was noted. The wearout intensity was calculated in terms of weight reducing in one revolution of the disk.

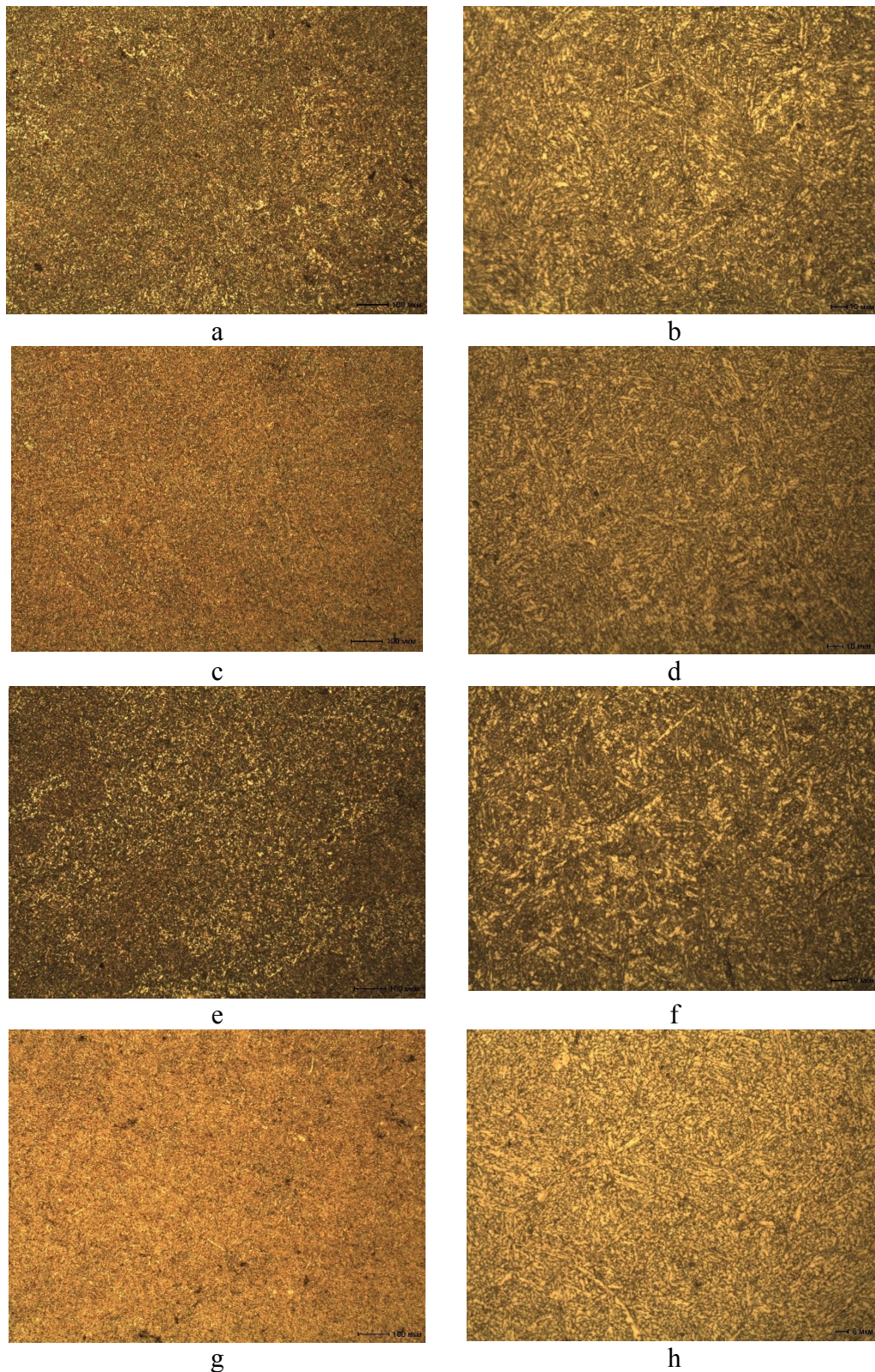
Metallographic analysis was conducted on optical microscope OLYMPUS GX-51 in the light field in the limits of range $\times 100-1000$ magnification after etching samples by nital to determine the degree of influence changes in chemical composition on parameters of the microstructure. Studying of longitudinal deposited metal samples on nonmetallic inclusions existence was conducted in according with GOST 1778-70 with $\times 100$ magnification. Size of former austenite grain was determined by GOST 5639-56 with $\times 100$ magnification. Martensite needles size was determined by GOST 8233-56 with $\times 1000$ magnification.

Results and Discussion

Hardness, wear intensity tests results and chemical composition analysis of the samples is shown in table 1.

Table 1 – Chemical composition, hardness, and wear intensity of the deposited metal.

Sample number	Components weight content %								Sample hardness, HRC	Sample wear intensity, grams per revolution * 10^{-4}
	C	Si	Mn	Cr	Ni	Mo	V	W		
1	0.23	0.12	0.85	1.45	0.53	0.51	0.68	0.08	25	1.65
2	0.21	0.23	0.89	1.45	0.54	0.55	0.54	0.03	21	1.15
3	0.17	0.18	0.85	1.40	0.52	0.54	0.63	0.05	21	1.11
4	0.17	0.28	0.91	1.32	0.45	0.46	0.59	0.06	21	1.12
5	0.15	0.20	0.79	0.08	0.56	0.25	0.51	3.38	17	1.15
6	0.14	0.17	0.75	0.07	0.5	0.13	0.47	2.57	16	4.08
7	0.14	0.26	0.78	0.09	0.55	0.68	0.56	2.88	16	1.70
8	0.12	0.21	0.71	0.07	0.52	0.5	0.44	2.43	15	1.79



a, b – sample №1; c, d – sample №2; e, f – sample №3; g, h – sample №4

Fig. 1 – Microstructure of the metal deposited by the flux cored wire system Fe-C-Si-Mn-Cr-Ni-Mo-V, (a, c, e, g $\times 100$), (b, d, f, h $\times 500$).

Metallographic analysis of the samples which were surfaced by flux cored wire system Fe-C-Si-Mn-Cr-Ni-Mo-V shown, that when carbon contain is 0.17-0.23% and chromium contain is within the limits 1.30-1.50% (samples №1-4) it provides uniform structure with fine-needed martensite (rating №3) which is formed inside the borders of the former austenite grain, retained austenite, which is presented in small amounts in the form of separate islands, and thin layers of δ -ferrite,

which is located on the borders of the former grain austenite (Fig. 1). Martensite needle size in the structure of the researched samples is in ranges 2-5 mcm (table 2). Size of former austenite grain conforms №6.

Use of CFCM as a carbon contain component provided high metallurgical quality of the surfaced metal. Nonmetallic inclusions impurity of all researched samples is inconspicuous: low amount of tiny nonmetallic inclusions was established, particularly non-deformable silicates and oxide spots (table 2).

Table 2 – Nonmetallic inclusions and structure characteristics of researched samples.

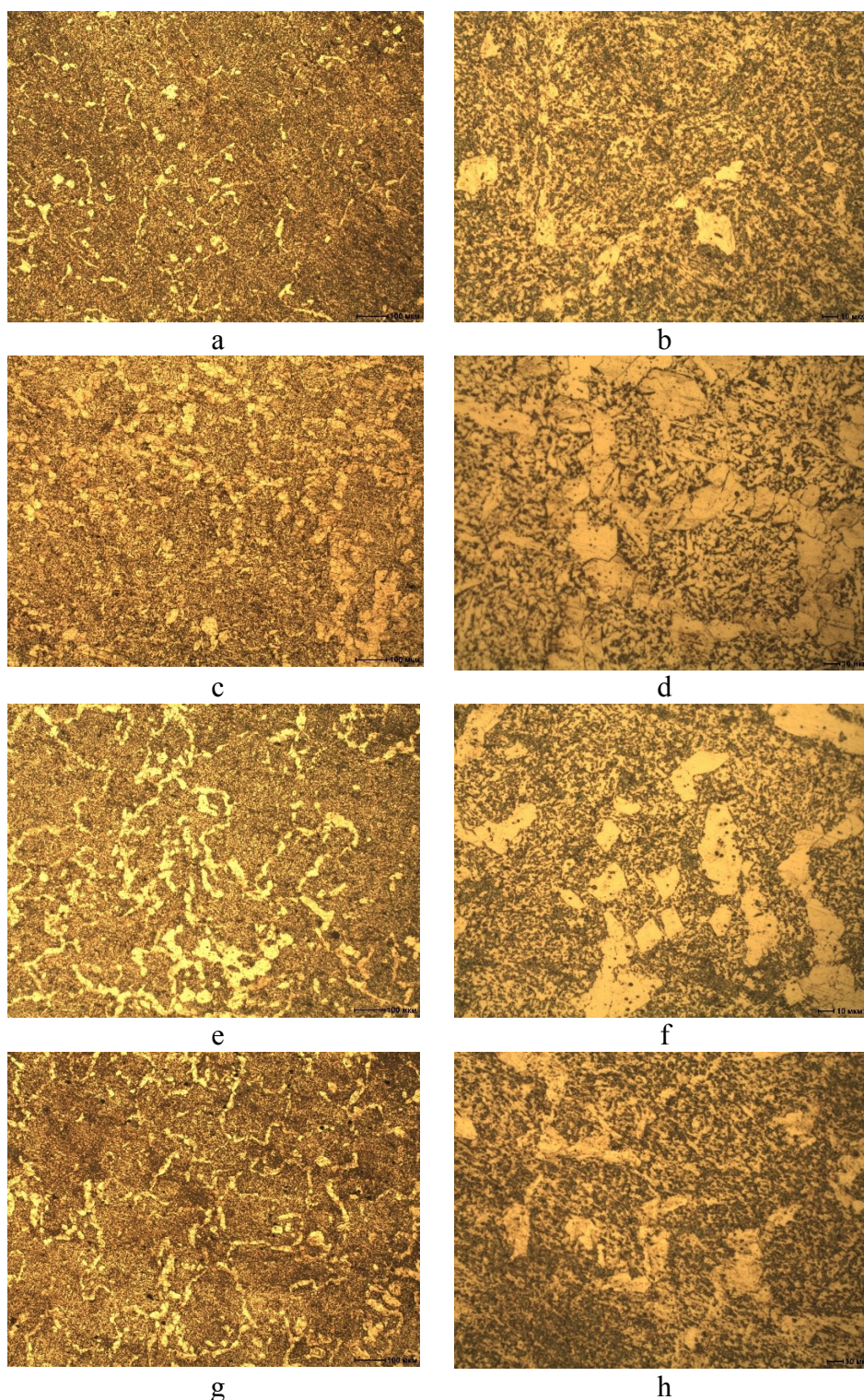
Sample number	The contamination of nonmetallic inclusions, rating		Size of austenite grain, rating	Size of martensite needles, mcm
	Non-deformable silicates (fragile)	Oxide spots		
1	1b, 2b, 3a	1a	6, 5	2-5
2	1b, 2b, 3a	1a, 2a	6	2-4
3	1b, 2b, 3a	1a	6	2-5
4	1b, 2b, 3a	1a	6	2-4
5	2b, 1b, 2a	1a	4, 5	-
6	2b, 1b, 2a	1a	4, 5	-
7	1b, 2b, 2a	1a	4	-
8	2b, 3b	1a	4	-

It is established, that chromium contains in the deposited layer in 1.32-1.45% amount, bars from growing austenite grain during surfacing processes and, consequently, promotes forming fine-needled martensite structure during metal cooling, that is confirmed by better rating of hardness and wear resistance in comparison with samples which are surfaced by the wire with higher tungsten content (table 1).

Metallographic analysis of metal surfaced by flux cored wire system Fe-C-Si-Mn-Ni-Mo-W-V shown, that appending tungsten in amount 2.43-3.38% while reducing carbon and chromium content down to 0.12-0.15% and 0.07-0.09%, consequently produce forming hypoppearlitic structure in the deposited layer (Fig. 2).

Samples №4-8 has hypoppearlitic structure with former austenite grain size №4, №5. What is more, there is big amount of retained austenite in the structure, that is located on boundaries of pearlite grain colonies. Also, is noticed presence of fine black inclusions in the boundaries of pearlite grain colonies, what is characteristically for tungsten carbides. Assigned structure changes determine reducing of deposited metal hardness and wear resistance (table 2).

With that said, established, that, despite tungsten carbides presence in hypoppearlitic structure of the samples, which was surfaced by low-carbon flux cored wire system Fe-C-Si-Mn-Ni-Mo-W-V, more suitable and preferable for surfacing equipment details and mechanisms, that works in abrasive wear conditions, is flux cored wire system Fe-C-Si-Mn-Cr-Ni-Mo-V. Apparently, hypoppearlitic matrix could not provide enough good counterstand against abrasive wear because of lower hardness, so formed tungsten carbides could not fully provide wear resistance, because they require more hard matrix for rigid adherence. Consequently, the conclusion is that hypoppearlitic structure is not preferable to incorporate it with tungsten carbides to increase wear resistance.



a, b – sample №5; c, d – sample №6; e, f – sample №7; g, h – sample №8

Fig. 2 – Microstructure of the metal deposited by the flux cored wire system Fe-C-Si-Mn-Ni-Mo-W-V, (a, c, e, $\times 100$), (b, d, f, h $\times 500$).

Conclusions

1. Established, that using lean alloyed flux cored wires of the systems Fe-C-Si-Mn-Cr-Ni-Mo-V and Fe-C-Si-Mn-Ni-Mo-W-V with the use as a carbon contain component CFCM provides forming deposited layers with low nonmetallic impurity level.
2. Using flux cored wire of the system Fe-C-Si-Mn-Cr-Ni-Mo-V for surfacing provides forming in the deposited layer uniform fine-neededled martensite structure with thin layers of δ -ferrite and

former austenite, that supply enough hardness and wear resistance for miner equipment details exploitation.

3. Using flux cored wire of the system Fe-C-Si-Mn-Cr-Ni-Mo-V provides obtaining deposited layers with lower carbon and chromium contain, but higher tungsten contain, that leads forming not martensite, but hypoppearlitic structure with tungsten carbides, which could not supply requiring hardness and wear resistance level.

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Development of Relay Control Systems of Power and Temperature Mode of Induction Crucible Furnaces with Use of Physical Modeling

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Keywords: model, temperature, melting, copper, magnesium.

Abstract. The physical models of the induction crucible furnaces with nonconducting crucible and conducting crucible were developed. Experimental study of the parameters of an induction crucible furnace for melting of ferromagnetic lumpy charge in ferromagnetic nonconducting crucible was made. Experimental study of the parameters of the furnace for melting copper and magnesium in conducting crucible was made. Three-position control system of active power of induction crucible furnace for melting of ferromagnetic lumpy charge and two-position control system for temperature regime of induction crucible furnaces with conductive crucible were developed.

1. Introduction

Induction crucible furnaces (ICF) are widely used in industry for melting ferrous and non-ferrous metals, in particular qualitative copper and magnesium alloys. Depending on electrical properties of material of the crucible distinguish ICF with the nonconducting and conducting crucible.

Furnaces with the dielectric ceramic crucible intended for melting metals belong to the first group. In such furnaces the charge (cage) heats up by the induced current in it, the crucible is equivalent to air gap.

Furnaces with the steel, graphite or clay-graphite crucible possessing bigger or smaller conductance belong to the second group. If crucible wall thickness more than twice exceeds penetration depth of current in crucible material, it is possible to consider that the induced current is concentrated in a crucible wall, the charge gets warm only by a heat transfer and can not possess conductance. At a smaller thickness of a wall of the crucible the electromagnetic field penetrates into the charge, and energy is emitted both in a crucible wall and in the charge if it is electrically conductive. Furnaces with the conductive crucible have heat insulation.

The modern units of ICF require automation of the technological process which is conducted to them, monitoring and control electrical (voltage, current, power, $\cos\varphi$) and technological (metal temperature) parameters. As the melting ICF in particular considered by ICF with the conductive crucible are rather inertial objects in the thermal relation, practice shows that the system of regulation by the thermal mode of the cage in furnaces can be built by the on-off principle. Identification of ICF as objects of control system with the subsequent embedding of model of the furnace in structural model of regulation system in MATLAB is necessary for a choice of parameters of regulation system of the furnace.

Identification of ICF can conducted with use of methods of electrical and thermal calculations or with use of physical modelling [1 – 3].

Physical modeling allows to get the most accurate information about the full-scale object. It is based on change of scale of physical properties of the materials, the geometrical sizes of system or parameters characterizing the heating mode with saving physical entity of the processes happening in the model and the original [2,4]. Using the theory of similarity, results of modeling can be transferred to furnaces of bigger capacity, applying scale factors.

2. Experimental Research of Physical Model of the ICF with Nonconducting Crucible for Ferromagnetic Lumpy Charge Melting

For experimental researches in the NRU "MPEI" was created a laboratory-industrial stand on the basis of ICF for melting copper, aluminum and magnesium, as well as ferromagnetic steel. Experimental studies were carried out for three metals: ferromagnetic steel, copper and magnesium.

Physical modeling on the ICF scale model was conducted during the heating of ferromagnetic lumpy furnace charge. Steel wire pieces 25 mm long, two different diameters (2.9 mm and 1.1 mm) were used to model this lumpy charge.

It is almost impossible to accurately reproduce in the model numerical values of all defining criteria in case of their large number and on condition of geometrical similarity.

In practice, as a rule, conditions of full similarity aren't met that forces to move to approximate simulation in case of which the model reproduces the same physical process as a real object, with partial violation of some of five conditions of full modeling [5].

The work was conducted on the model of ICF unit with a ceramic crucible. The furnace is powered from the transistor sequential inverter with a maximum rated capacity of 2.5 kW and with a frequency of 16 – 21 kHz [3].

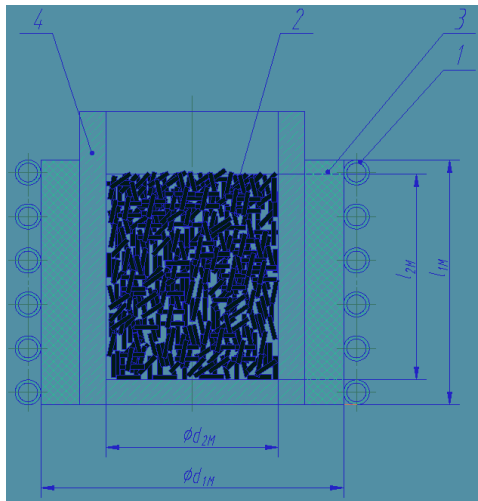


Figure 1. Drawing of ICF physical model: 1 – inductor; 2 – lumpy charge; 3 – heat insulation; 4 – refractory crucible

Figure 1 shows a sketch of a modelled ICF. Parameters of the model unit are following: $d_{1m} = 94$ mm, $d_{2m} = 67$ mm, $l_{1m} = 76$ mm, $l_{2m} = 113$ mm, weight of the charge $m_{cm} = 985$ g, frequency at the initial moment of heating of furnace charge $f_m = 16.8$ kHz, the number of turns of the inductor – 6.

Parameters of the model unit of ICF can be led to the parameters of the big ICF according to the theory of similarity. For geometrical similarity we have:

$$\frac{d_{1f}}{d_{1m}} = \frac{d_{2f}}{d_{2m}} = \frac{l_{1f}}{l_{1m}} = \frac{l_{2f}}{l_{2m}} = \frac{\sqrt{f_m}}{\sqrt{f_f}} = \frac{d_{fc}}{d_{cm}} = \sqrt[3]{\frac{m_{fc}}{m_{cm}}}$$

where d_{1f} , d_{2f} , l_{1f} , l_{2f} , f_f , d_{fc} , m_{fc} — an inside diameter of the inductor, charge diameter, inductor height, charge height, supply frequency, furnace charge diameter, weight of the furnace charge respectively for the big furnace, d_{cm} — charge diameter in the model unit. The provided model can be compared with ICF for melting

steel with the following characteristics: $d_{1f} = 940$ mm, $d_{2f} = 670$ mm, $l_{1f} = 760$ mm, $l_{2f} = 1130$ mm, frequency $f_f = 170$ Hz, capacity of the furnace m_{fc} is 1 ton.

During the experiment the values of current of the inductor I , voltage of the inductor U , frequency of power voltage f , active power consumed by the frequency converter P and average temperature in the charge T (fig. 2) were measured. Additional parameters were also measured in addition to key parameters: water temperatures in the input and output of the inductor, the flow of the cooling water of the inductor using the rotameter with an impulse output, temperatures of a crucible wall, the cover and the hearth of the furnace. The additional parameters were used for indirect determination of thermal losses of ICF and average temperature of the furnace [1,2,4]. PLC (programmable logic controller) «Oven PLC 150» with 4 built-in ADC modules (analog-to-digital converter) and 8-channel analog input module «Oven MB110-8I» which is connected to the PLC by the sequential asynchronous communication channel of the RS-485 standard according to the Modbus RTU protocol were used for measurement and registration of analog signals. The PLC calculates average temperature by indirect parameters and regulates energy supply in the furnace charge by on-off regulation.

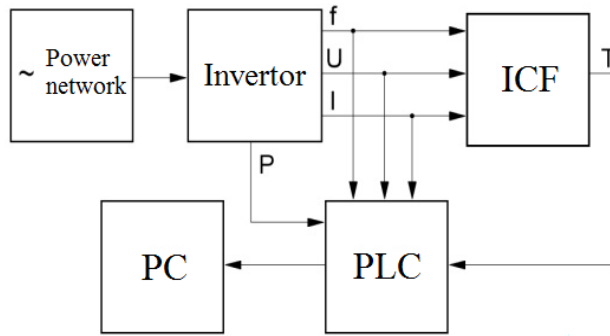


Figure 2. Laboratory unit functional scheme

The measured ICF parameters were registered in a file on the personal computer (PC) with a 1 second period using the established system of automated data collection based on the PLC. The received parameter values were averaged with 30 seconds period. Then there is processing of the received array of experimental data with the subsequent analysis of results.

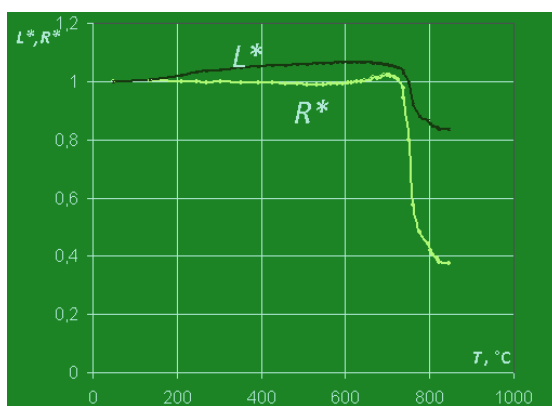
The ICF equivalent circuit parameters R_i and L_i (active resistance and inductance of “inductor - charge” system) are calculated

from the received data. Definition L_i instead of corresponding inductive resistance X_i is made for the reason that a supply frequency automatically adapts under the changing of the charge during heating process.

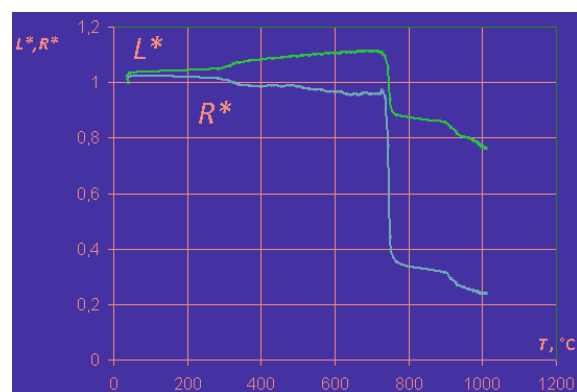
Calculated dependences of the parameters of the ICF serial equivalent circuit are graphically presented in Fig. 3a and 3b with two different diameters of lumpy furnace charge made from ferromagnetic steel in the form of temperature dependencies.

Figure 3 shows relative units of R^*_i and L^*_i parameter values of sequential electrical equivalent circuit of «inductor – charge» system, which are referred to the values in the beginning of the experiment, and in the case a) initial values are $R_i = 0.057$ Ohm, $L_i = 4$ μ H, in the case b) initial values are $R_i = 0.063$ Ohm, $L_i = 4$ μ H. The ratio of the field penetration to the diameter of the charge at the initial moment of heating for the case of a) $\Delta/d_{cm} = 0.17$, b) $\Delta/d_{cm} = 0.06$ for the charge with diameters of 1.1 mm and 2.9 mm [2].

Figure 3 shows that in the process of the charge heating parameters of ferromagnetic lumpy charge are strongly changing. Under such circumstances in order to maintain the constancy of the power in the charge during the heating it is necessary to regulate the voltage of the inductor, which is usually realized using a frequency converter with self-tuning parameters for ICF. Using dependences of the change of the charge parameters, it is possible to select the power supply of necessary power [6, 7]. The mathematical description of the ITP obtained as a result of experimental research is used in the synthesis of power and temperature regulators constructed according to the relay principle [8, 9].



a)



b)

Figure 3. Dependence of the relative values of impedance R^* and total inductance L^* of the inductor on temperature for the diameters of the charge: a - 2.9 mm; b - 1.1 mm

3. Control System of the Energy Efficient Frequency Converter for Induction Crucible Furnace for Melting of Ferromagnetic Lumpy Charge

To heat and melt the metal in the ICF, the power supplied to the charge during the entire process should be maintained unchanged. This measure allows you to melt the load in minimum time. This saves energy. To do this, the ICF power supply should change the voltage on the inductor, keeping the value of the power supplied to the load, maximum and unchanged. Consider the case where the power supply of the ICF is a thyristor frequency converter (TFC).

The TFC scheme shown in Fig. 4, has wide possibilities to regulate output parameters due to the possibility of rebuilding its topology.

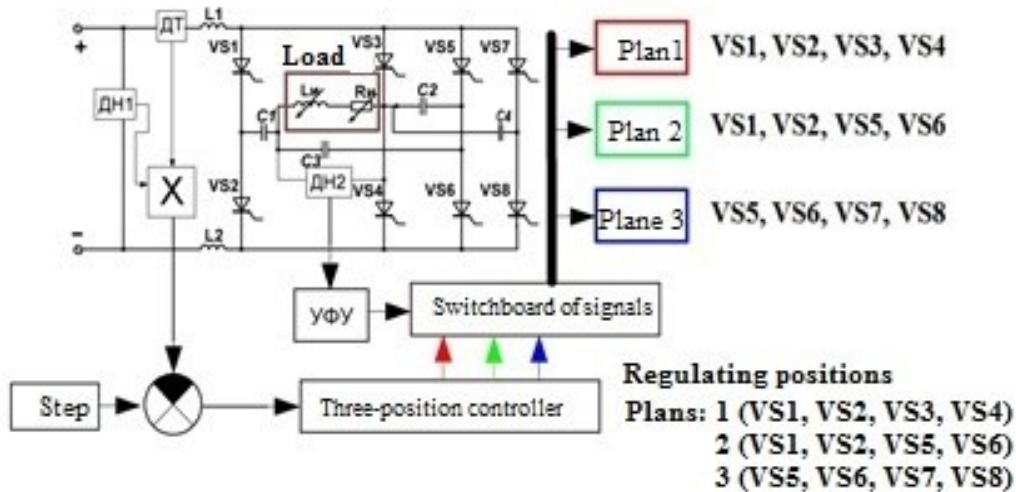


Figure 4. Structural TFC control circuit

The considered TFC has three different topologies (Fig. 4). It allows the implementation of relay three-position regulation of the active power of the ICF. The parameters of the frequency converter are controlled by the active power consumed by the frequency converter. Thus, the tunable TFC with three-position control will be switched either between circuit 1 and circuit 2, or - between circuit 2 and circuit 3. The average value of the active power in the load will correspond to the setpoint power of the three-position controller [10 – 13].

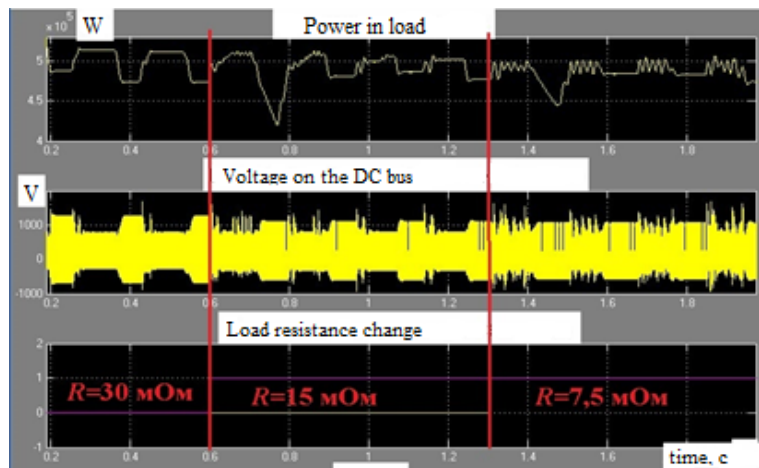


Figure 5. Power control diagrams with varying load resistance

During the simulation, the average power was measured when the control system changed the resistance R in the range $7.5 \div 30 \text{ m}\Omega$ at $L = 25 \text{ }\mu\text{H}$, both with decreasing resistance R and with increasing it. The power control diagrams for changing the load resistance are shown in Fig. 5. It follows from the experimental results that the static control accuracy is no worse than $\pm 5\%$ when the load parameters change ($R = 7.5 \div 30 \text{ m}\Omega$ and $L = 20 \div 40 \text{ }\mu\text{H}$).

4. Experimental Research of Physical Model of the ICF with Conducting Crucible and Development of Control System

For melting copper and magnesium, conductive crucibles of graphite and ferromagnetic steel are used, respectively. When conducting experimental studies to regulate the value of the input voltage, a single-phase laboratory autotransformer Fnex with a power of 3 kVA is used. To control the temperature, PLC "Oven PLC 150" is used. The functional diagram of the laboratory bench is shown in Fig. 6.

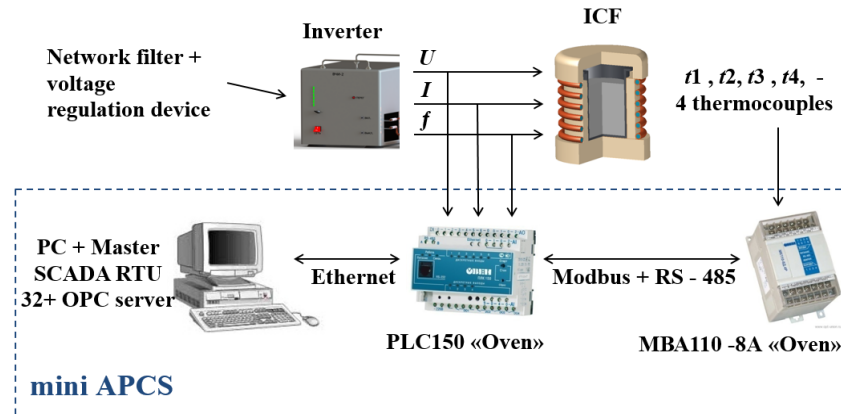


Figure 6. Functional diagram of laboratory installation

Figure 7 shows a drawing of «inductor – charge» system furnace with a conductive crucible with geometric sizes.

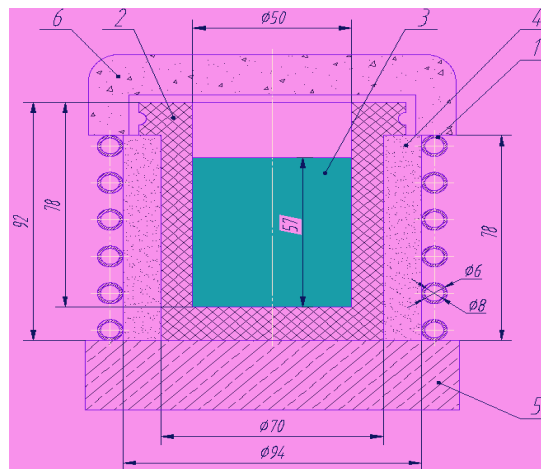


Figure 7. Drawing of «inductor – charge» system furnace with lining: 1 – inductor; 2 – conductive crucible; 3 – metal; 4 – ceramic insert; 5 – hearth; 6 – cover

Using the experimentally obtained mathematical description of ITP for melting copper in the MATLAB package, a structural model of a two-position system for controlling the temperature of the charge in the furnace for the heating, melting and holding of copper in the furnace has been developed (Figure 8) [3]. ICF with a conductive crucible is represented by a first-order inertia link with a gain factor of 0.8 and a time constant of 620 s (Transfer Fcn1 block). The range of temperature fluctuations is 10 °C and is set in the Relay block. The temperature setpoint is 1200°C (Step1 block). The feedback of the control system includes a temperature sensor (thermocouple), represented by a first-order inertia link with a time constant of 15 s (Transfer Fcn).

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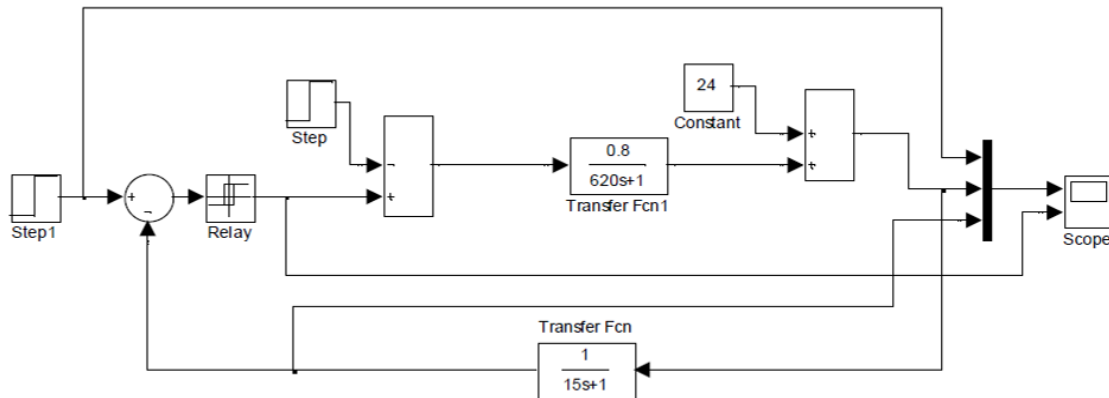


Figure 8. Structural model of a two-position system for regulating the temperature of the ICF in the mode of heating, melting and holding a copper melt

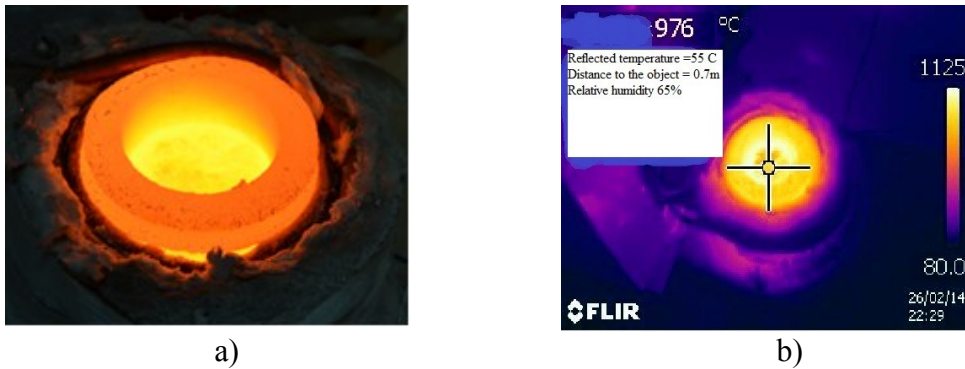


Figure 9. Photo (a) and a thermovision image (b) of the crucible with a liquid melt

Figure 9 shows a photo and thermal image of the crucible with liquid copper. Figure 9a corresponds to the preceding metal drain. The initial loading of the copper charge was made in the cold crucible, then after a while in process of melting and upsetting the charge two additional charges were made. The total mass of melted copper was about 1 kg, the general time of melting - about 19 min, overheating temperature – 1120°C . A convex meniscus 3 - 4 mm high was observed after the complete melting.

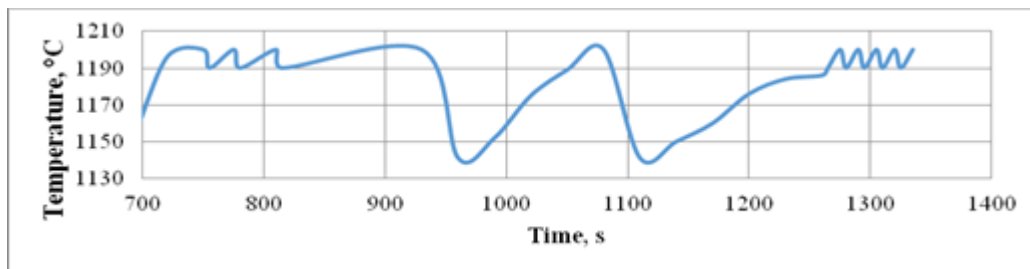


Figure 10. Graph of temperature versus time during melting of copper charge

Figure 10 shows an experimental curve of temperature change taking into account adding of copper charge in melted metal and the value of the oscillation range. The temperature oscillation range in quasi steady state in process of regulation is from 1190 to 1200°C . The system is completely fulfills all the disturbances, the main of which is adding of the charge in the molten metal. The obtained data with an accuracy of at least 5% coincide with the theoretical curve obtained in MATLAB using the structural model of Figure 8.

A similar structural model was developed for ICF for the melting of magnesium, and transient processes of temperature and power change were obtained. The experimental investigation of the

regulator consisted in removing the curve of the change in the temperature of the metal in the furnace.

Using the developed model of the control system, it is possible to configure a two-position system for adjusting the temperature of the ITP cage, depending on the capacity of the furnace, the power of the power source, the level of the driving force, the changes in the perturbations, and others.

5. Conclusion

Based on the results of studies of the control system model of the tunable TFC in Simulink, the accuracy of regulation is no worse than $\pm 5\%$ in a wide range of load variations with the control system parameters set out above.

The use of tunable TFC, controlled by the method of three-position control, allows to reduce losses in the converter unit, reducing the power consumption during the entire process of melting steel in the ICF.

A structural model of a two-position regulator of the temperature of the ICF cage with a conducting crucible for melting copper and magnesium is obtained. The results of adjusting the controller are given. The developed models of the temperature regulator of the ICF cage with a conductive crucible have been verified using an experimental setup. Differences in the temperature of the charge do not exceed 5%, which is quite acceptable.

Since melting ICF are sufficiently inertial objects (the thermal time constant of the crucible, as a rule, is more than 300 s), the control system for the thermal regime of the cage can be built on a two-position principle.

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On the Shape of the Boundary of Adsorption Lines of Molecules under the Limited Volume

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Keywords: boundary of the adsorption line, collision against the walls, influence of the cell flexion.

Abstract. The authors complete calculations of the boundary of adsorption lines of CO and CO₂ molecules under the limited volume. The cells with the transverse size from 10 to 200 nm are considered. The inaccuracy of the half-breadth of the lines appearing when the cylindrical and spherical cells are modeled with rectangular cells was estimated. When calculating the shifts and the half-breadth of the lines of CO and CO₂ molecules the authors consider the model in which CO and CO₂ molecules adsorbed on the walls of the pore may cease their rotational motion.

Introduction

Study of nanopore materials is essential as they are applied in many fields, such as catalysis, gas storing, optics, biomedicine, etc. [1]. A number of papers are devoted to experimental and theoretical study of the absorption spectrum of molecular gas (CO, CO₂, O₂, H₂O, NH₃) under the conditions of cells of varying sizes and chemical composition (SiO₂, Al₂O₃, ZrO₂ and TiO₂) (see, for example, [2-7] and references to them). In the given papers the authors show that in the cells of small (nanoscale) sizes collisions of molecules against the walls of the cells and molecular collisions are the determining mechanisms for the broadening of vibrational-rotational lines of absorption. At the same time there are no papers where the authors analyze the influence of the surface of the cell upon the vibrational-rotational spectrum of gas absorption. Influence of the surface upon the vibrational spectrum of the molecules which are chemically and physically adsorbed in it can be significant and it is well-studied experimentally [8-10]. Physically adsorbed molecules can continue rotational motion (for example, adsorption of H₂ on the surface Cu) or may lose it. In [11] the authors suppose that physically adsorbed molecules of water in the nanopores of aerogel may influence upon the adsorption spectrum of water vapor contained in the nanopores.

The given paper is devoted to studying the shape of adsorption lines of CO and CO₂ molecules situated in the cells with transverse size from 10 to 200 nm. The paper comprises two parts. In the first part the authors calculate the contribution of collisions of molecules against the walls of the pore into the half-breadth of molecule lines. The authors make quantitative estimation of approximations caused by modeling cylindrical and spherical cells with rectangular cells, i.e. flexion of the cell is taken into account.

In the second part of the paper the authors calculate the contribution of collisions of CO and CO₂ molecules against each other and against the molecules adsorbed on the walls of the pore into the half-breadth of the lines of CO and CO₂ molecules. The case when the molecules lose their rotational motion is considered. Then, the combined contribution of the mechanisms of broadening associated with intermolecular collisions and collisions against the walls of the cells into the total boundary of the line is calculated.

Collisions against the Walls of the Cells

Boundary of adsorption line can be calculated as the real part of Fourier transformation of the averaged autocorrelation function $\Phi(t)$ [12].

$$I(\omega) = \frac{1}{\pi} \operatorname{Re} \int_0^{\infty} \Phi(t) \exp(i\omega t) dt, \quad (1)$$

where rotational frequency $\omega = 2\pi c\nu$, where c is the velocity of light and ν (cm^{-1}) – wave number. Broadwise the lines make varying contributions [13] the most important of which may be Doppler broadening, collision broadening (broadening due to the collisions between the molecules) and broadening due to the collision of molecules against the walls. The least studied mechanism of broadening is that resulting from the collision of molecules against the walls, making its contribution Γ_{Wall} into the total half-breadth of the lines Γ_{tot} .

Value Γ_{Wall} can be estimated according to formula [14]

$$\Gamma_{\text{Wall}} (\text{cm}^{-1}) = \frac{v_{\text{av}} A}{2\pi c 4V} \quad (2)$$

where $V_{\text{av}} = 4(kB T / 2 \pi m)^{1/2}$ is mean molecular velocity, kB – Boltzmann's constant, A and V determine the area and the volume of the cell.

For more proper estimation of this contribution we need to know the autocorrelation function from (1). This function is determined as the probability of the molecule not having any collisions against the walls during the time interval t and, according to [3, 4] can be put down as follows

$$F_{\text{Wall}}(t) = \frac{1}{V} \int_V dx dy dz \int_{-x/t}^{x/t} dv_x \int_{-y/t}^{y/t} dv_y \int_{-z/t}^{z/t} dv_z f(v_x, v_y, v_z) e^{-t\alpha(v)} \quad (3)$$

Here $f(v_x, v_y, v_z) = 1/[(v_0\pi^{1/2})^3] \exp[-(v^2/v_0^2)]$ – function of molecular velocity distribution v , $v^2 = v_x^2 + v_y^2 + v_z^2$, $v_0 = (2kB T / m)$ – the most probable velocity and $\alpha(v)$ is the probability of molecular collisions with the walls in unit time. For the case $\alpha(v)d/v_0 \ll 1$ (d – average transverse size of the cell) the exponent in (3) can be neglected, then

$$F_{\text{Wall}}(t) = \frac{1}{V} \int_V dx dy dz \int_{-x/t}^{x/t} dv_x \int_{-y/t}^{y/t} dv_y \int_{-z/t}^{z/t} dv_z f(v_x, v_y, v_z) \quad (4)$$

If the cell is modeled with a rectangular cell ($2L = d_1$) $\times$ ($2M = d_2$) $\times$ ($2N = d_3$), then integral (4) is the product of 3 integrals

$$F_{\text{Wall}}(t) = F_{\text{Wall}}(L, t) \cdot F_{\text{Wall}}(M, t) \cdot F_{\text{Wall}}(N, t), \quad (5)$$

each of them determines a univariate correlation function, for example

$$F_{\text{Wall}}(L, t) = L^{-1} \int_0^{2L} dx \int_0^{x/t} f(v_x) dv_x \quad (6)$$

where $f(v_x) = 1/(v_0\pi^{1/2}) \exp[-(v_x/v_0)^2]$. Integral (6) is calculated [3, 4] and produces

$$F_{\text{Wall}}(L, t) = \Phi(d/v_0 t) - t v_0 / (d\sqrt{\pi}) \{1 - \exp[-(d/v_0 t)^2]\} \quad (7)$$

where

$$\Phi(u) = \frac{2}{\sqrt{\pi}} \int_0^u e^{-u^2} du \quad (8)$$

is the integral of probabilities [15] (in [16] it is denoted by $\operatorname{erf}(u)$). Integral (6) determines the correlation function for the cell which is modeled by two infinite plates situated at the distance of $d = 2L$ from each other. Function $F_{\text{Wall}}(t)$ (5) is real so integral (1) with function $\Phi(t) = F_{\text{Wall}}(t)$, can be put down as follows

$$I(\nu) = \frac{1}{\pi} \int_0^{\infty} F_{\text{Wall}}(t) \cos[2\pi c(\nu - \nu_0)t] dt \quad (9)$$

If the cell is modeled by parallelepiped $d_1 \times d_1 \times d_2$, the in (9) $F_{Wall}(t)$ is determined as

$$F_{Wall}^{Bar} = [F_{Wall}(d_1/2, t)]^2 F_{Wall}(d_2/2, t) \quad (10)$$

For the cubic cell $d_1 \times d_1 \times d_1$ in formula (9)

$$F_{Wall}^{Cube} = [F_{Wall}(d_1/2, t)]^3 \quad (11)$$

For the cylindrical cell with height d_2 , with the circular disk $d_1 = 2R$ in diameter as the base, integral (4) is reduced to the product of two integrals,

$$F_{Wall}^{Cylinder} = F_{1-2}(R, t) \cdot F_{Wall}(d_2/2, t) \quad (12)$$

where integral $F_{Wall}(d_2/2, t)$ is still determined by formula (6) and the double integral $F_{1-2}(R, t)$ is determined by the expression

$$F_{1-2}(R, t) = \frac{1}{S_R} \iint_D \Phi(x/v_0 t) \cdot \Phi(y/v_0 t) dx dy, \quad (13)$$

where D is the domain of integration, which boundary is circumference $(x-R)^2 + (y-R)^2 = R^2$, and square $S_R = \pi R^2$. When domain D is replaced by the square with the side $d_1 = 2R$ integral (13) gives $[F_{Wall}(d_2/2, t)]^2$, so (12) goes into (10). For the spherical cell $d_1 = 2R$ in diameter integral (4) is reduced to triple integral

$$F_{Wall}^{Sphere} = \frac{1}{V_R} \iiint_V \Phi(x/v_0 t) \cdot \Phi(y/v_0 t) \cdot \Phi(z/v_0 t) dx dy dz \quad (14)$$

Modeling of the spherical cell with a cubic cell with the side $d_1 = 2R$ brings us to formula (11). The difference in the results of calculations of half-breadth of Γ_{Wall} lines of the water vapor molecule completed with application of the simplified formula (2), on the one hand, and according to formula (9) with application of (8), on the other, is shown in [11]. This difference becomes most apparent for the cells of small sizes of 10-50 nm.

Results of Calculations

Calculation of the boundary of the lines $I(\nu)$ was completed according to formula (9) where function $F_{Wall}(t)$ was taken in four different types (10) – (14) corresponding to the cells of different shape. The sizes of the cell varied from 10 to 200 nm. The results of the calculations depend upon the average molecule velocity v_{av} in the cell, upon its size and do not depend upon frequency ν_0 . For molecule of CO frequency $\nu_0 = 2139.39 \text{ cm}^{-1}$ was chosen as ν_0 : it corresponds to the frequency of transition $J = 1 \rightarrow J - 1$ in the range 1-0. The calculated boundaries are well described by Lorentz curve

$$I(\nu) = \frac{1}{\pi} \frac{\Gamma}{(\nu - \nu_0)^2 + \Gamma^2} \quad (15)$$

From fitting to this curve the half-breadth of the line was determined, in this case half-breadth Γ_{Wall} resulting from collisions of the molecules against the walls of the vessel.

The results of calculations for CO molecule are reduced to the following.

1). When modeling the cell as a parallelepiped with the square side $10 \text{ nm} \leq d_1 \leq 200 \text{ nm}$ in the base and height $d_2 = 10d_1$ maximal relation $\Gamma_{Wall}^{Bar} / \Gamma_{Wall}^{Cylinder}$ of the half-breadth of the line calculated for the cell in the shape of a parallelepiped (Γ_{Wall}^{Bar}) to the half-breadth of the line calculated for the cylinder ($\Gamma_{Wall}^{Cylinder}$) made 1.05 for $d_1 = 10 \text{ nm}$, i.e. maximal inaccuracy in the calculation of the half-breadth of the line associated with the replacement of the cylindrical cell with a rectangular one in the calculations made 5%. With increase of d_1 this inaccuracy reduces and for the cells with $d_1 > 80$ it makes less than 1%.

In Fig.1 the boundaries of lines $I(\nu)$ (9) calculated for the parallelepiped and the cylinder of the same height d_2 are shown. The boundary of the base of cylinder is the circumference d_1 in diameter inscribed into the square with the side d_1 which is the base of the parallelepiped. Height d_2 was chosen so that $d_2/d_1 = 10.0$.

2). When modeling the cell as a cue with the side of $10 \text{ nm} \leq d_1 \leq 200 \text{ nm}$ maximal relation $\Gamma_{Wall}^{Cube} / \Gamma_{Wall}^{Sphere}$ of the half-breadth of the line calculated for the cell in the shape of a cube (Γ_{Wall}^{Cube}) to the half-breadth of the line calculated for the sphere (Γ_{Wall}^{Sphere}), made 1.19 for $d_1 = 10 \text{ nm}$, i.e. maximal inaccuracy in calculation of the half-breadth of the line associated with replacement of the spherical cell with a cubic one inscribed into a sphere in the calculations made 19% for the cell with the smallest edge $d_1 = 10 \text{ nm}$. To compare with, this inaccuracy made 9%, 5% and 1% for $d_1 = 50$, 80 and 200 nm accordingly.

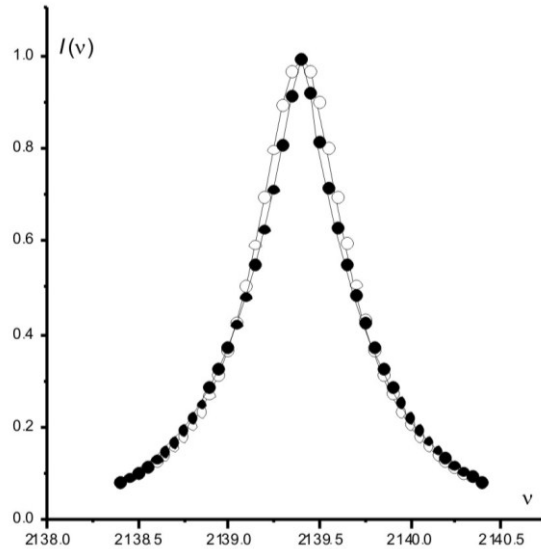


Fig. 1. The boundary of the line of CO molecule calculated according to formula (9) for a cylinder (light symbols) and for parallelepiped (dark symbols) of the same height $d_2 = 10 d_1$; $d_1 = 10 \text{ nm}$ – diameter of the circle and side of the square lying in the base of the cylinder and parallelepiped accordingly.

In Figure 2 the boundaries of the lines $I(v)$ (9) calculated for the cubic cell with the edge of $d_1 = 10 \text{ nm}$ and for the spherical cell with diameter $d_1 = 10 \text{ nm}$ are presented.

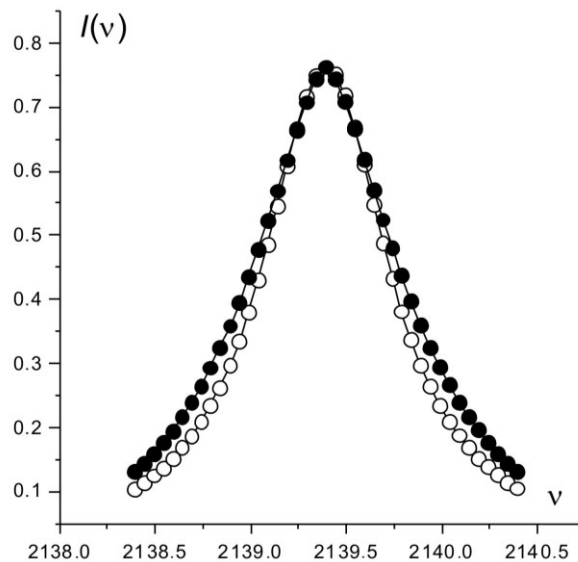


Fig. 2. Boundary of the line of molecule CO calculated according to formula (9) for a spherical cell (light symbols) with radius $d_1 = 10 \text{ nm}$ and for a cubic cell (dark symbols) with the edge $d_1 = 10 \text{ nm}$.

Similar calculations of inaccuracies for the molecule of CO_2 lead to practically similar conclusions.

Dependence of the half-breadth of the lines Γ_{Wall}^{Bar} and Γ_{Wall}^{Cube} of CO and CO_2 upon the size of the cells is determined in [17].

Intermolecular Collisions

The total half-breadth of the lines of molecules Γ_{tot} under the conditions of the cell, beside the contribution Γ_{wall} , is determined by the contribution Γ_{if} (mol), resulting from the intermolecular collisions and, probably, (as it is noted in [11]) with contribution $\Gamma_{if(ad)}$ resulting from inelastic collisions of absorbing molecules against the molecules adsorbed on the surface of the cell (CO) ' (or (CO₂)). The adsorbed molecules marked with the prime may have differing (in comparison to free molecules) electro-optical parameters or may lose their rotational motion. It is determined by the forces acting between the surface and the molecules [8-10].

To calculate the contribution Γ_{if} (mol) = P · γ_{if} and shift Δ_{if} (mol) = P · δ_{if} (P – pressure) we used various approximations for calculating the broadening γ and shift δ ratios. In Corf-Leavitt approximation (CL) [18]

$$\gamma_{if} + i\delta_{if} = \frac{n}{c} \sum_{J_2} \rho_{J_2} \int_0^\infty v F(v) dv \int_0^\infty b db [1 - (1 - S_2^{middle}) \exp(-iS_1 - S_{2,i}^{outer} - S_{2,f}^{outer*})] \quad (16)$$

Here n – concentration of broadening molecules, ρ_{J_2} – statistical weight of level J_2 of these molecules, $F(v)$ – velocity v distribution of molecules. All other distributions are determined in [18]. In Robert-Bonamy approximation (RB) [19]

$$\gamma_{if} + i\delta_{if} = \frac{n}{c} \sum_{J_2} \rho_{J_2} \int_0^\infty v F(v) dv \int_0^\infty b db [1 - (1 - S_2^{middle'}) \exp(-iS_1 - S_{2,i}^{outer} - S_{2,f}^{outer*} - S_2^{middle''})] \quad (17)$$

Here $S_2^{middle''}$ is part $S_2^{middle'}$, which is diagonal along J_2 and $S_2^{middle'} = S_2^{middle} - S_2^{middle''}$.

When applying the kinematic model from [19] for the distance $r_c(b)$ of the closest approximation between the interacting molecules the inner integral is (16), (17) is changing according to

$$\int_0^\infty b db \rightarrow \int_{r_0}^\infty D \cdot r_c dr_c, \text{ where } D \text{ is transition Jacobian from the variables } (b, v) \text{ to } (r, v), \text{ the inferior}$$

limit of the integral r_0 depends upon the parameters of the isotropic interaction potential of ϵ and σ . Transition to $r_c(b)$ also supposes application of parabolic motion paths for the interacting molecules. Thus, under the chosen intermolecular potential of interaction, there are 4 possible variants of calculation of broadening and shift ratios. In the paper we choose the following potential as the potential of interaction

$$V = \{V^{elect}\} + \{V^{pol}\} = \{V_{\mu-\mu} + V_{\mu-q} + V_{q-\mu} + V_{q-q}\} + \{V^{pol}\} \quad (18)$$

consisting of electrostatic and polarizing parts. In the electrostatic part the following interactions are taken into account: dipole-dipole ($\mu-\mu$), dipole-quadrupole ($\mu-q$), quadrupole-dipole ($q-\mu$) and quadrupole-quadrupole ($q-q$). In the polarizing part V^{pol} , only far-reaching forces are taken into account corresponding to contribution r^{-6} (r – the distance between the centers of mass of the interacting molecules). All contributions into the interrupt function $S_2(b)$ corresponding to potential (18) are described in [18]. The values for $\mu = 0.1$ D, $q = -2.5$ DA, polarizability $\alpha = 1.94 \text{ \AA}^3$ for CO and similarly $q = -4.3$ D·Å, $\alpha = 2.59 \text{ \AA}^3$ for molecule of CO₂ as well as harmonic vibrational frequencies, anharmonicity constants and rotational constants of molecules were taken from [20].

In Fig. 3 we present the results of the calculations of the half-breadth of the lines Γ_{if} (mol) for P branch ($J \rightarrow J+1$, $m = -J$) of the rotational band 0-0 of CO molecule calculated according to formula (16) (line 1) with application of kinematic model for $r_c(b)$ and parabolic paths (parameters $\epsilon = 100.2$ K and $\sigma = 3.763$ Å were taken from [19]), according formulas (16) (line 2) and (17) (line 3), accordingly, with application of linear paths. The results are reduced to pressure $P = 1$ atm. To compare with, dark symbols in this figure present calculations of Γ_{if} (mol) completed with

application of empiric formula for the broadening ratios $\gamma_{CO-CO} = \sum_{n=0}^4 b_n |m|^n$ which parameters b_n

(they are presented in [13], for example) were obtained from analysis of experimental data for the self-broadening ratios of the lines of band 0-3 of CO molecule [21]. It is apparent that there is difference in the asymptotical behavior of the calculated ratios γ .

In Fig.4 we compare the half-breadths of the lines $\Gamma_{if}(mol)$ of P branch ($J \rightarrow J+1$) of CO_2 molecule (self-broadening, $P=1$ atm.), calculated in [22] (line 1) and according to (17) (line 2). In [22] the calculation was completed according to RB method [19]; in the intermolecular potential the electrostatic contribution and short-range interaction forces are taken into account. From the figure we can see good agreement between the values of $\Gamma_{if}(mol)$ calculated according to different methods. Deviations in the values of Γ_{if} can be also associated with different choice of quadrupole moment (in [19] for CO_2 $q = -3.9D \cdot A$ is given).

Further calculations were completed according to formulas (16)-(17) which do not require knowing the parameters of isotropic interaction potential.

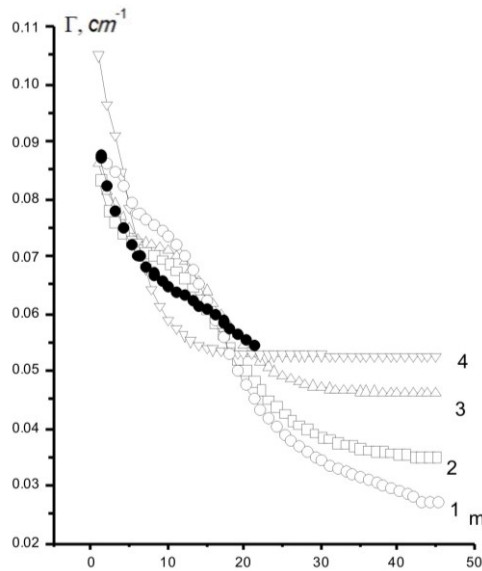


Fig. 3. Collisional half-breadths $\Gamma_{if}(mol)$ of rotational lines of CO molecule (self-broadening) calculated according to formula (17) with consideration to the kinematic model from [19] (parabolic paths, line 1), according to formula (17) (line 2), according to formula (16) (line 3), $T=296$ K, $P=1$ atm. Line of dark symbols was obtained from processing experimental data [21]. Line 4 determines half-breadth $\Gamma_{if}(ad)$ of the same lines of CO molecule but for the case of their broadening with non-rotating CO molecules.

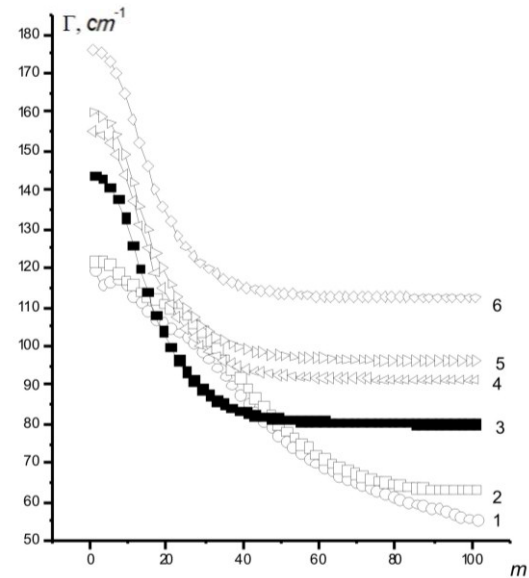


Fig. 4. The collisional half-breadths of the rotational lines $\Gamma_{if}(mol)$ of CO_2 molecule, calculated according to formula (16) (line 2) and according to Robert-Bonamy method in [22] (line 1). Line 3 determines half-breadth $\Gamma_{if}(ad)$ of the same lines of CO_2 molecule but for the case of their broadening with non-rotational CO_2 molecules. Lines 4, 5 and 6 determine the total half-breadth of the lines of the molecule $\Gamma_{tot} = \Gamma_{Wall} + \Gamma_{if}(mol)$ in the cells 30, 20 and 10 nm accordingly.

Collisions against the Molecules Adsorbed on the Walls

The hypothesis about the possibility of broadening the lines of free molecules in the cell by physically adsorbed molecules on the walls is presented in [11]. The prerequisites of such hypothesis are also given in [9]. The calculation of broadening and shift of the adsorption lines of water vapor molecules adsorbed on the walls of the nanopore by non-rotating (H_2O) molecules is in reasonable agreement with the experimental data from [6]. It was revealed that broadening and shift by non-rotating molecules (H_2O)' significantly differ from broadening and shift of the lines by rotating molecules. This is particularly so with the shift of the lines. Let us notice, although, that molecule of H_2O in contrast to CO and CO_2 molecules has greater dipole moment and such broadening can be specific for the molecule and for the given surface. As well as in [11] calculation of the half-breadth $\Gamma_{if}(ad)$ and shift $\Delta_{if}(ad)$ was completed with application of formulas (16), (17). In formulas (16), (17) n now means concentration of physically adsorbed molecules of CO' or (CO_2)'. To keep the dimensionality this concentration was related to the unit of volume. In [11]

concentration of n was considered as free parameter. In the current paper it is settled towards the value $n = 2.6868 \cdot 10^{19} (T_0/T) \text{ cm}^{-3}$ which corresponds to concentration n of free molecules, $T_0 =$ The parameters of the intermolecular potential of interaction chosen were the same as those chosen when calculating $\Gamma_{if}(\text{mol})$. Thus, the adsorbed molecule of $(\text{CO})'$ or $(\text{CO}_2)'$ was presented as molecule of $(\text{CO})'$ ($(\text{CO}_2)'$) without the rotational structure. In Fig. 3 and 4 the calculated values of half-breadths of the lines $\Gamma_{if}(\text{ad})$ of CO and CO_2 molecules are shown for the case of their broadening by non-rotating molecules of $(\text{CO})'$ or $(\text{CO}_2)'$ ($P = 1 \text{ atm.}$). We observe absolutely different dependence of contributions $\Gamma_{if}(\text{ad})$ upon $|m| = J$. For the same lines shifts Δ_{if} were calculated. For the rotational transitions the function of first-order transmittance $S_1(b) = 0$ in formulas (16), (17) so shifts Δ_{if} were completely determined by the imaginary components of resonance functions. The results of the calculations are shown in Fig. 5 for the lines of CO_2 molecule; for the lines of CO molecule similar dependence of Δ_{if} on $|m|$ was obtained. From the figure we see that the shifts of the lines determined by the collisions against the non-rotating molecules significantly differ from the shifts of these lines determined by the collisions against free molecules.

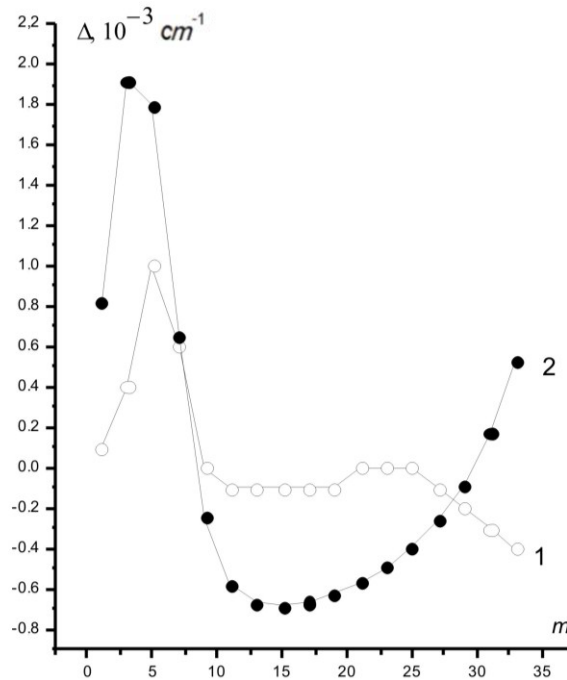


Fig. 5. Shifts Δ_{if} of the rotational lines of CO_2 molecule caused by pressure of its own (line 1) and collisions against the adsorbed molecules (line 2).

Joint Consideration of Intermolecular Collisions and Collisions against the Walls

Let us study the joint consideration in the boundary of the lines of intermolecular collisions and collisions against the walls which occur under any type of cell surface. Under their joint consideration the correlation function $\Phi(t)$ in integral (1) should be put down as the total

$$\Phi(t) = F_{\text{Wall}}(t) \cdot F_{\text{mol}}(t), \quad (19)$$

where function $F_{\text{mol}}(t)$ is correlation function for the intermolecular collisions. It describes the possibility that the molecule will not collide against other molecules during time t . As well as in (4) this function was applied in the form

$$F_{\text{mol}}(t) = \exp(-2\pi c \Gamma_{if}(\text{mol}) t), \quad (20)$$

where $\Gamma_{if}(\text{mol})$ determines the collisional (between free molecules) breadth of the line. In Fig. 6 the boundaries of the line $J_i = 0 \rightarrow J_f = 1$ of the vibration band 1-0 of CO molecule calculated for the case of its self-broadening, broadening due to collisions against the walls of spherical cell $d = 10 \text{ nm}$ in diameter and the common boundary calculated according to formula (1) with general correlation

function (19). For the considered cell the common boundary is determined mainly by the collisions against its walls. In the considered model of calculations $\Gamma_{tot} = \Gamma_{Wall} + \Gamma_{if}$ (mol) in this connection $\Gamma_{Wall} = 0.4 \text{ cm}^{-1}$, $\Gamma_{if} = 0.083 \text{ cm}^{-1}$. Contribution Γ_{Wall} remains unchanged for all lines and contribution Γ_{if} (mol) changes. In Fig. 7 the results of similar calculations for the cell 80 nm in diameter are shown. For this cell constant contribution Γ_{Wall} into the general breadth of the lines makes 0.05 cm^{-1} . As the diameter of the cell d increases this contribution reduces rapidly [17].

The calculations for CO_2 molecule lead to similar results. Constant contribution $\Gamma_{Wall}(\text{CO}_2)$ for the lines of this molecule is smaller than contribution $\Gamma_{Wall}(\text{CO})$ for the lines of CO molecule, in this connection $\Gamma_{Wall}(\text{CO}) / \Gamma_{Wall}(\text{CO}_2) = [(m(\text{CO}_2)/m(\text{CO}))^{1/2}]$ for practically all cells under consideration (m denotes the mass of molecule) which corresponds to (2). Dependence of the general half-breadth of lines Γ_{tot} on $|m| = J$ upon the size of the cell is shown in Fig. 4 (lines 4, 5 and 6).

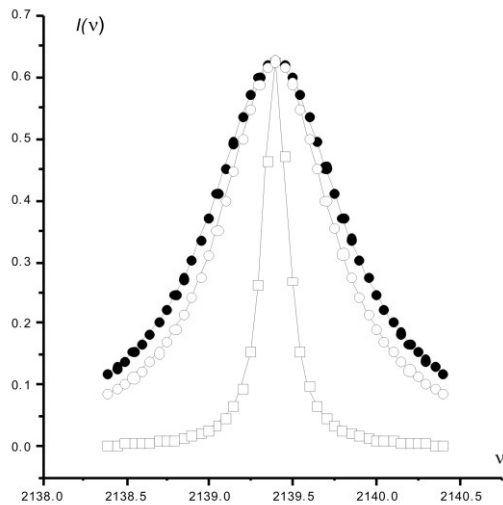


Fig. 6. The boundaries of the line of CO molecule with center $\nu_0 = 2139.39 \text{ cm}^{-1}$ calculated for the case of its self-broadening (light squares, $T = 296 \text{ K}$, $P = 1 \text{ atm}$), broadening with the walls of the spherical 10 nm in diameter (light circles) and total boundary calculated with the correlation function (16) (dark circles).

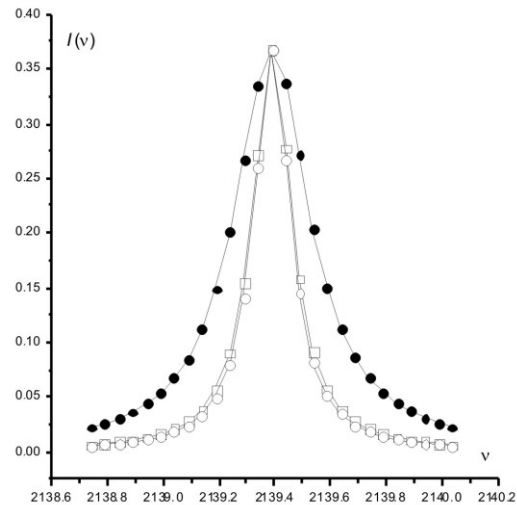


Fig. 7. The same that in Fig. 6 but for the cell with $d = 80 \text{ nm}$.

Discussion

In the paper for the adsorption lines of CO and CO_2 molecules in the cells with transverse size from 10 to 200 nm the authors show that boundaries of lines are well described by Lorentz model boundary. The authors analyze contributions Γ_{Wall} and Γ_{if} (mol) into the total half-breadth of the lines Γ_{tot} , caused by the collisions of molecules against the walls of the cell and intermolecular collisions under room temperature and pressure $P = 1 \text{ atm}$. The deal of each contribution into the total half-breadth of the lines depends upon the size of the cell where the absorbing gas is contained. For the cells with transverse size of 10 nm the dominating contribution is contribution Γ_{Wall} which is determined by collisions against the walls. It makes 80% of the total half-breadth of line Γ_{tot} . For the cells with transverse size $\sim 50 \text{ nm}$ the contributions of two types of collisions are compatible, for the cells with transverse size $\sim 200 \text{ nm}$ the deal of Γ_{if} from intermolecular collisions into the total half-breadth Γ_{tot} becomes over 80%. The authors also consider the specific variant of broadening the lines of CO and CO_2 molecules – broadening with non-rotating adsorbed molecules of CO' and $(\text{CO}_2)'$. The possibility of such variant realization can not be known beforehand.

Modeling of cylindrical and spherical cells with rectangular cells introduces certain inaccuracies into the calculated values for half-breadth of the lines. These inaccuracies are most significant for cells ~ 10 nm and reach 19% when modeling a spherical cell with a cubic one with the edge equal the cell diameter. As the size of the cell increases contribution Γ_{wall} into the total half-breadth of the line reduces as well as the inaccuracy introduced by the replacement of cylindrical or spherical cell by their rectangular models. For the cells with transverse sizes 200 nm the inaccuracy in the calculations resulting from such replacement is less than 1%.

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Effect of Electropulsing on the Fatigue Behavior and Change in the Austenite Steel Structure

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Abstract. The tests were carried out to identify the influence of electropulse treatment on austenite steel ((mass %) 0.44C, 16.50Mn, 0.26Cr, 0.08Ni, 0.34Si, 2.74Al, 0.002S, 0.017P, Fe – balance). The fatigue life is reported to increase by 1.8 times. Electron microscopic research into the dislocation structure of the steel was conducted under diverse fatigue conditions with the purpose to give reasons for the identified effect. The dislocation chaos substructure, reticular and fragmented dislocation substructures were found in the steel in the initial state. Fatiguing leads to the change in the dislocation substructure parameters. The subsequent electropulse treatment furthers transformation of the grain structure since grains arise and grow due to evolving local dynamic recrystallization and partial transformation of the dislocation substructure and occurrence of a great number of microtwins. The increase in the fatigue life is associated with the mentioned above transformations resulting from electropulse treatment of the steel structural state.

Introduction

The authors of a lot of scientific papers have dealt with fatigue failures of metals and alloys. However, this issue is still a challenging one. The results known in the scientific world emphasize in most cases the complicated character of the problem in spite of offering any ways of its solution. The physical nature of fatigue is associated with self-organization, accumulation and interaction of lattice defects in the process of fatiguing [1-4]. It is also difficult to reveal stages of accumulated fatigue failures, especially those ones which are caused by requirements for geometry of samples and their initial state. Measuring of ultrasound speed is one of the frequently used methods to monitor stages of fatigue micro-failures [5]. The main point of this method is in plotting the speed of ultrasound transmission against the number of fatigue cycles of the material to be tested. The dramatically changing curve which presents the relation of ultrasound speed to the number of fatigue cycles demonstrates that a critical structural state originates in the material, and the development of this state is the reason for the further failure of the material [5-9].

Electropulse treatment helps stop failure origination and increase the fatigue life of steels up to 25%. The point of it is that powerful electric pulses with the specified parameters pass through the material subject to preliminary fatigue deformation [10-18].

The data on regularities of electropulse impact on the dislocation substructure and phase composition of the materials to be processed are necessary for the development and practical application of this method. A number of studies were carried out by our scientific group to reveal the influence of electropulse and electron beam treatment on various structure classes of steels [6, 7, 19, 20]. The article presents how the above problem was considered for austenitic wear resistant steel via analyzing the evolution of the defect substructure and phase composition of steel.

Material and Methods of Research

Austenitic steel containing (mass %) 0.44C, 16.50Mn, 0.26Cr, 0.08Ni, 0.34Si, 2.74Al, 0.002S, 0.017P, Fe – balance was used as the material for experiments. In the process of samples preparation for fatigue tests the long axis of the sample was parallel to the rolling direction. The

fatigue tests were carried out using the facilities operated on the base of cyclic symmetric console bending. Sizes and form of the samples were similar to those given in [7]. All samples had concentrators in form of a cut out semicircle with the diameter of 10 mm. The cut out was made on one side of the sample. The tests were carried out at the ambient temperature. The frequency of bending was 18 Hz. In conditions of normal stresses samples got destructed after the testing sequence $N_1=102\ 000$. Samples were not tested after the testing sequence $N_2=70\ 000$. Some selected samples were subject to electropulse treatment (the frequency of pulses $f=19\text{ Hz}$) for 45 seconds. Then, the fatigue tests were made till the samples got destructed. It was found out that destruction of the samples required the testing sequence $N_3=179\ 000$.

Some billets were cut off the samples to carry out diffraction electron microscopy tests using the device EM-125. They were polished and subject to electrolytic thinning for making foils. The material was in four states for the purpose of research: 1) the initial state (before tests); 2) in the mid-phase of fatiguing; 3) after electric treatment in the mid-phase of fatiguing; 4) the destructed one at multi-cyclic fatigue; 5) the destructed one at multi-cyclic fatigue after electric treatment. At least five samples in each state were used for the purpose of research. Then the data were averaged and statistically processed.

The structural state of the steel was analyzed in sections perpendicular to the longitudinal ax-is of the sample 0.5, 0.8 and 2.3 mm from the surface of destruction, as well as in the zone border-ing closely the surface of sample destruction.

Results and their Discussion

The dislocation structure of the steel in the initial state comprises the dislocation chaos substructure (proportion by volume $P_v=10\%$) (Fig. 1a), non-disoriented reticular ($P_v=7\%$), disoriented reticular with abrupt low-angle boundaries ($P_v=11\%$) (Fig. 1b) and fragmented ($P_v=72\%$) (Fig. 1 c) dislocation substructures. Besides the above dislocation substructures there are grains (25-30% of the total number of grains) with a few dislocations.

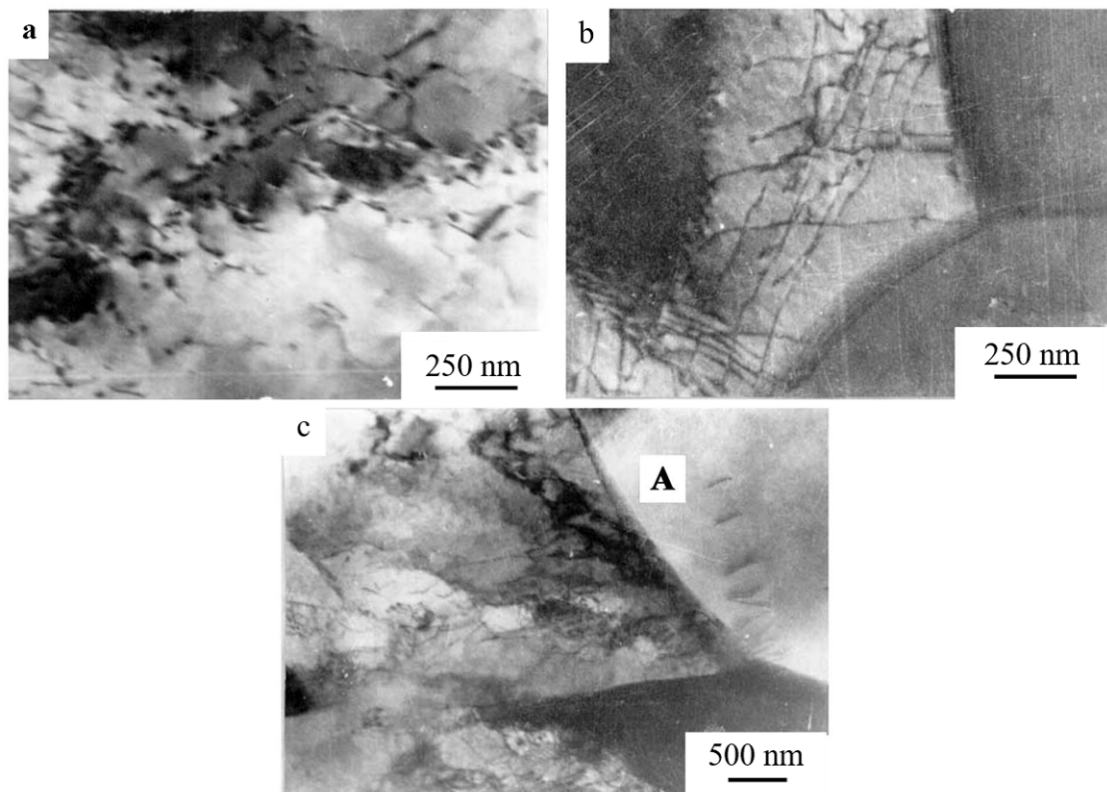


Fig. 1. Types of dislocation substructures in austenitic steel in the initial state: a) dislocation chaos, b) reticular, c, d) fragmented. The letter A points at the grain with a low scalar density of dislocations.

Furthermore, zones with ϵ -martensite crystals are found in the steel. These zones are located along the grain boundaries making it possible to assume the reason of martensite $\gamma \rightarrow \epsilon$ transformations: changing chemical composition of the steel because alloying elements move towards the grain boundaries and elastic stresses arising close to the boundaries of the adjacent grains.

As identified by the numerical analysis of the dislocation substructure and stress state of the steel in the initial state, the scalar density of dislocations in the reticular dislocation substructure is $\rho = 1.32 \cdot 10^{10} \text{ cm}^{-2}$, and the bending rotating amplitude of the crystalline lattice is $\chi = 394 \text{ cm}^{-1}$, whereas those ones in the fragmented substructure are $\rho = 0.52 \cdot 10^{10} \text{ cm}^{-2}$ and $\chi = 510 \text{ cm}^{-1}$, respectively. This fact demonstrates that the fragmented dislocation substructure is the most strained structural component of the steel in the initial state.

Fatiguing of steel causes the transformation of the dislocation chaos substructure into the reticular dislocation substructure (Fig. 2). The proportion by volume of the fragmented substructure does not change in this case. Deformation of the steel is associated in the reticular substructure with the increase in the scalar density of dislocations up to $\rho = 1.81 \cdot 10^{10} \text{ cm}^{-2}$, and the rotating bending amplitude of the crystalline lattice up to $\chi = 518 \text{ cm}^{-1}$. The scalar density of dislocations rises up to $\rho = 0.52 \cdot 10^{10} \text{ cm}^{-2}$ in the fragmented substructure, whereas the rotating bending amplitude is reduced up to $1.09 \cdot 10^{10} \text{ cm}^{-2}$.

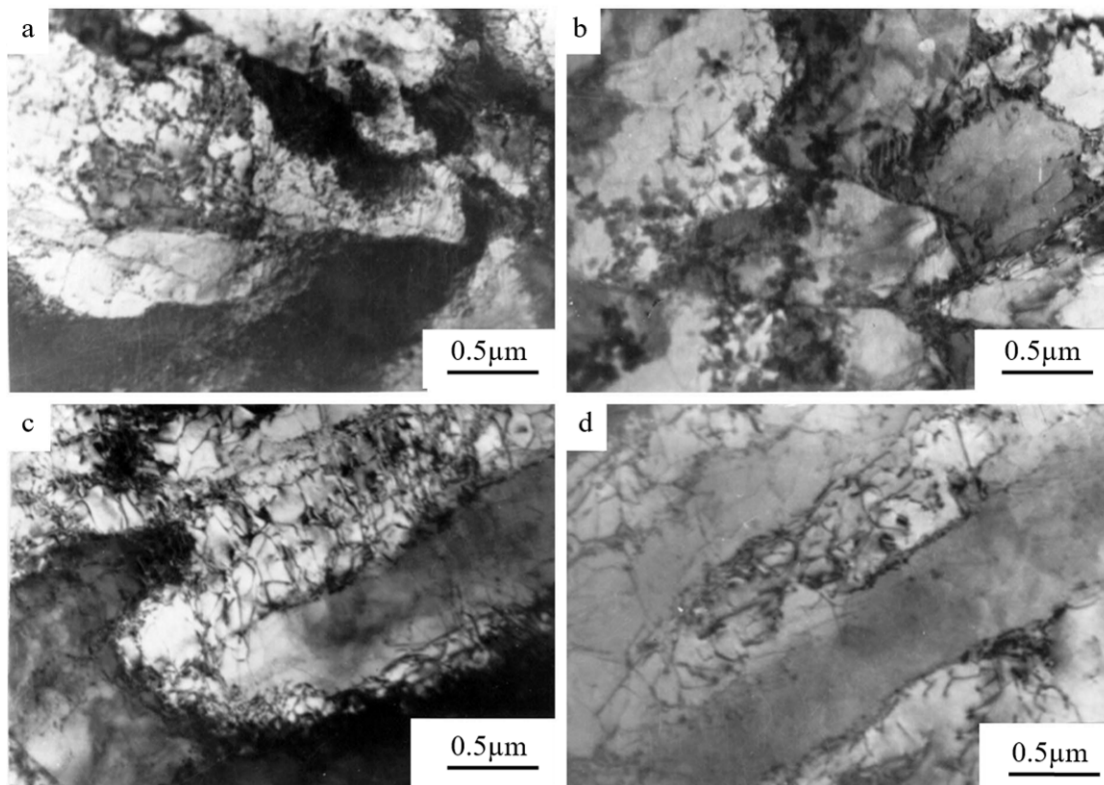


Fig. 2. Electron-microscopic image of the dislocation substructure forming in austenite steel when fatigue testing up to 70 000 cycles: a) reticular dislocation substructure; b) fragmented dislocation substructure; c, d) anisotropic fragments a, b) zone of the material subject to maximal fatiguing; c, d) $\sim 0.8 \text{ mm}$ from the maximally fatigued zone of the material.

The increase in the scalar density of the dislocations in the reticular substructure results in formation of weakly disoriented bands. These bands (Fig. 2 c, d) are anisotropic fragments with the dislocation structure inside them. In the structure of the steel after fatigue tests there are a lot of deformation micro-twins.

The types of dislocation substructures are not changed when fatiguing up to destruction. The proportion by volume of chaotically distributed dislocations is 18%, that one of reticular dislocation substructure – 42%, and fragmented substructure – 40%.

As shown by comparison of the obtained results, the dislocation chaos structure formed in the initial state of steel disappears at some stages of fatigue tests. However, the dislocation chaos structure was revealed in the steel when further testing. Therefore, a conclusion can be made that this dislocation substructure arises when steel fatiguing.

So, the fatigue deformation of steel is the reason for the following transformations in the dislocation substructure: dislocation chaos - reticular – fragmented dislocation substructures.

Fatiguing is associated with formation of micro-twins in the zone of destruction; these twins are formed in the reticular dislocation substructure (Fig. 3). However, twins were not found at 70 000 testing cycles. Therefore, a conclusion can be drawn on the base of the compared numerical data that the twins arise because of the heightened amplitude of long-ranging stresses (it is proportional to rotating bending of the crystalline lattice) which is higher in the reticular dislocation substructure.

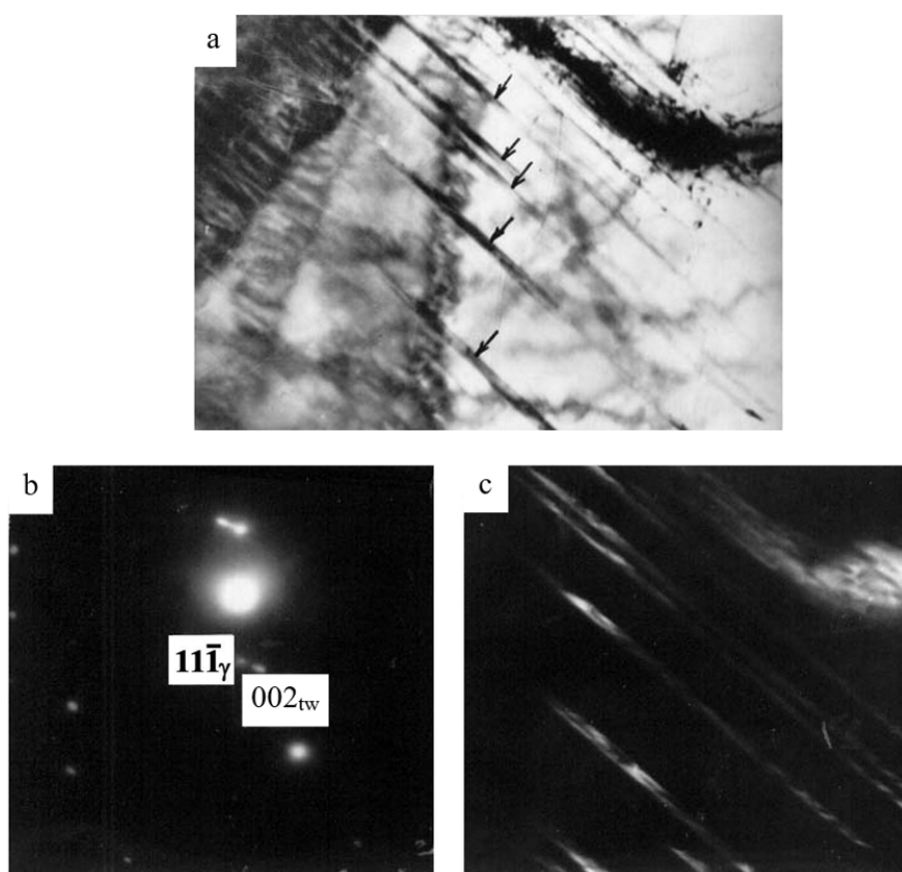


Fig. 3. Micro-twins (shown by arrows), forming in the process of electropulse treatment in austenitic steel: a – bright-field image, b – dark field in the undertone [002] of a twin (the undertone of the dark field is given), c – micro-electron-diffraction pattern of the foil zone.

The number of ϵ -martensite crystals is nearly constant at 70 000 testing cycles, although their proportion by volume increases significantly (almost by 2.5 times) in the zone of destruction. The reticular dislocation substructure is the main zone where ϵ -martensite crystals are formed.

Electropulse treatment of samples at the mid-phase of experiments (70 000 cycles of fatigue testing) is the reason for the following processes in the material. The grain structure is transformed – grains arise and grow due to the evolving local dynamic crystallization. Both fragmentation and recrystallization are registered. The number of dislocations is also reduced; as a consequence, the chaos dislocation structure occurs. There is a partial rearrangement of the dislocation substructure: the proportion by volume of the reticular dislocation substructure decreases ($P_v=38\%$), the proportions by volume of the chaos ($P_v=7\%$) and fragmented substructures ($P_v=55\%$) increase. A great number of micro-twins are formed and found mainly in the reticular dislocation substructure.

The main type of the dislocation substructure in the zone of steel destruction (after preliminary electropulse treatment at the mid-phase of fatiguing) is the reticular dislocation substructure, its

proportion by volume increases significantly with the growing number of testing cycles ($P_v=77\%$). The drop of proportion by volume of the dislocation chaos and the fragmented dislocation substructure is registered ($P_v=4$ and 19%). It should be mentioned that it is the reticular dislocation substructure where ϵ -martensite occurs more frequently and micro-cracks arise on interphase boundaries and inside them. This fact demonstrates that the reticular dislocation substructure causes steel destruction.

As shown by the SEM-results, there is a change in the electron structure of the solid solution and its concentration composition. Alongside with the micro-twins their seed dislocations are found (Fig. 3c). Electropulse treatment reduces the concentration of ϵ -phase, resulting, therefore, in augmentation of efforts for formation of micro-cracks. Furthermore, the electron structure of steel changes due to the alteration of concentration in the solid solution, it is one of the reasons for improving fatigue life of steel alongside with the varying long-ranging stress fields.

Conclusions

It is emphasized in the paper that the number of cycles needed for destruction increases 102000 to 179000 due to electropulse treatment when accumulating fatigue damages. Using the diffraction SEM-methods two types of dislocation substructures were found in austenitic steel: the dislocation chaos substructure, as well the reticular and fragmented substructures. As a result of steel fatiguing there is a transformation of the dislocation substructure. The further electropulse treatment is the reason for grain structure transformations since grains arise and grow due to the evolving local dynamic recrystallization and partial change in the dislocation substructure, as well as because of a great number of micro-twins. The reason for the increased fatigue life is in the above transformations caused by electropulse treatment of the steel structural state.

Acknowledgements

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Electron-Beam Welding - Structural-Phase State and Microhardness in the Weld Zone in a Submicrocrystalline Titanium Alloy Grade2

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Keywords: electron-beam, welding, structural-phase, microhardness.

Abstract. The paper presents the results of the X-ray diffraction analysis of structural-phase states in the weld zone of a titanium alloy Grade2 in micro- and submicrocrystalline states. It is established that the structural-phase state in the weld zone and in the heat-affected zone depends on the state of samples of the alloy Grade2 before welding. It is shown that formation process of metastable phases ω -Ti and α'' -Ti occurs in the submicrocrystalline state in the alloy Grade2 in the weld zone and in the heat-affected zone. Investigations of the features of the microhardness distribution in the weld zone in alloys Grade2 in micro- and submicrocrystalline states are carried out. Different character of microhardness distributions in the weld zone in the samples depending on the structural-phase state of welded plates made of alloy Grade2 is determined.

Introduction

Today, an increase in the strength and plasticity of metals and alloys by grinding grains, fragments (or subgrains), and substructure cells is widely used. In this connection, titanium alloys transferred from micro-, submicro-, and nanocrystalline states using intense plastic deformation methods are promising [1-4]. On the one hand, the stability of alloys in the submicrocrystalline state is a very important issue in connection with the expanding practical applications. On the other hand, the features of the structure and properties of alloys will be able to determine new properties of treated surfaces, their joints, and coatings. Therefore, it is important to know the laws of formation of structural-phase states when connecting parts from materials in the submicrocrystalline state by welding. This has been a pressing issue for a long time. To solve the problem, it is necessary to have experimental data on features of the change in structural-phase states in the weld zone.

The aim of this work is to study the structural-phase states of welded joints, obtained using the electron-beam welding method, of a titanium alloy Grade2 in micro- and submicrocrystalline state using the X-ray structural analysis method.

Materials and Techniques

Titanium alloy Grade2 was chosen for the study. The chemical composition of this type of alloy is shown in Table 1.

Table 1. Chemical composition of titanium alloy Grade2

Element	Fe	N	C	H	O	Other impurities	Ti
Content, weight %	0.25	0.04	0.07	0.01	0.2	0.3	remainder

The samples of the titanium alloy of the submicrocrystalline structure were prepared by rolling bars with a cross section of 6×6 mm, obtained by one of the methods of an intense plastic deformation (abc-pressing) [5]. The samples for welding consisted of plates 2×12 mm with a length of 30 mm. The samples were pre-cleaned and then clamped in a special tool to ensure free placement of the weld zone from the contact with the welding equipment in a vacuum chamber of the industrial installation ELS-0.5-6 located at the JSC SPC “Polyus”. Electron-beam welding of the samples was carried out using a gun with a plasma cathode [6], accelerating voltage (V) of 28 kV, beam current (In) of 35 mA, focusing-coil current of up to 551 mA, and beam travel speed of 25 mm/s.

The phase composition of initial samples and welds was carried out using an X-ray diffractometer DRON4-07 modified to digital signal processing. Filming was carried out on CuK α according to the Bragg-Brentano scheme. The quantitative phase analysis was carried out using the Rietveld method based on processing of diffractograms of the alloy Grade2 [7].

The microhardness was determined on cross sections of the samples in different sections of the weld using an automatic microhardness tester “Instron Wilson Hardness Tukon 2500-6” with a load of 25 – 50 g. The exposure time for the indentation was 5 s. Microhardness measurements were made along 5 parallel “tracks” in the plane of transverse sections perpendicular to the weld. The distance between the tracks was 200 μm .

Results and Discussion

X-ray diffraction study data

Figures 1 and 2 show the diffractograms obtained from samples of the titanium alloy Grade2 in micro- and submicrocrystalline states in the weld zone and away from the weld. The analysis of diffractograms of the initial alloy Grade2 in the microcrystalline state has shown that the alloy contains only one α -Ti phase. Since, on the diffractograms of the samples of the Grade2 alloy, only clear reflexes from the α -Ti phase are detected (Fig. 1, a).

The analysis of diffractograms of the alloy Grade2 in the microcrystalline state in the weld zone has made it possible to reveal that, in addition to the α -Ti phase, titanium oxides are also present in the weld zone (Fig. 1, b).

The diffractograms taken from the samples in the submicrocrystalline state of the alloy Grade2, obtained as a result of a step-by-step grinding of grains in the samples using the abc-pressing method with a parallel stepwise temperature decrease in the range of 750-500 °C, are shown in Fig. 2, b. It can be seen that the diffractograms of the samples in the submicrocrystalline state of the alloy Grade2 exhibit different ratio between the intensities of reflexes of the main phase α -Ti. This reflects the phenomenon of texture formation in the process of obtaining a submicrocrystalline state. Moreover, additional reflexes were found on the diffractogram (Fig. 2, a). These reflexes belong to the β -Ti phase. Thus, the X-ray diffraction method has revealed that in this state the alloy Grade2 is in a two-phase state: ((α -Ti) + (β -Ti)). Using the quantitative X-ray phase analysis method, it has been established that in this state the investigated alloy contains about 85% of α -Ti and 15% of β -Ti.

Fig. 2, b shows diffractograms of the alloy Grade2 in the submicrocrystalline state from a local location in the weld zone and in the heat-affected zone. In the given diffractogram, weak reflexes from phases α' -Ti, ω -Ti and titanium oxides have been found in addition to clearly observed reflexes Ti and β -Ti, (Fig. 2, b). This indicates that exposure of the alloy Grade2 in the submicrocrystalline state to an electron beam has led to a change in the structural-phase state. This is caused by the fact that formation of metastable phases α' -Ti and ω -Ti has occurred at the intersection in the region $\alpha \leftrightarrow \beta$ of the phase transition at high cooling rates after the end of welding.

The obtained data of X-ray diffraction studies reflect the ability of titanium alloys to form metastable phases as a result of the mechanical and thermal treatment [8]. According to the published data, the weld zone and the heat-affected zone in titanium alloys have a different number of characteristic sections that reflect different limiting heating temperatures and different cooling rates [9]. The presence of a temperature gradient and different cooling rates in different local regions in the weld zone results in a structural-kinetic step-like transformation $\beta \rightarrow \alpha$ in titanium alloys.

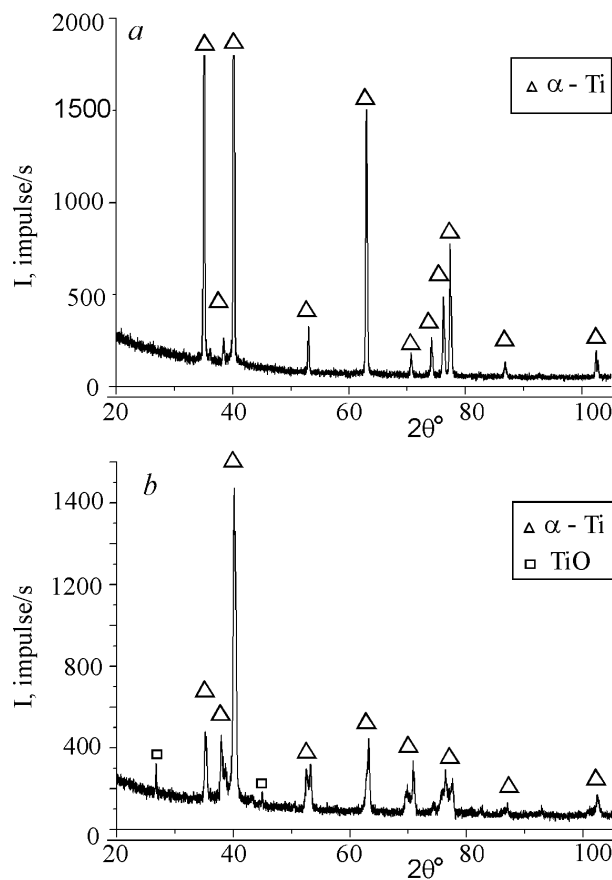


Fig. 1. Diffraction patterns of the alloy Grade2 in the microcrystalline state: a is away from the weld zone; b is in the weld zone

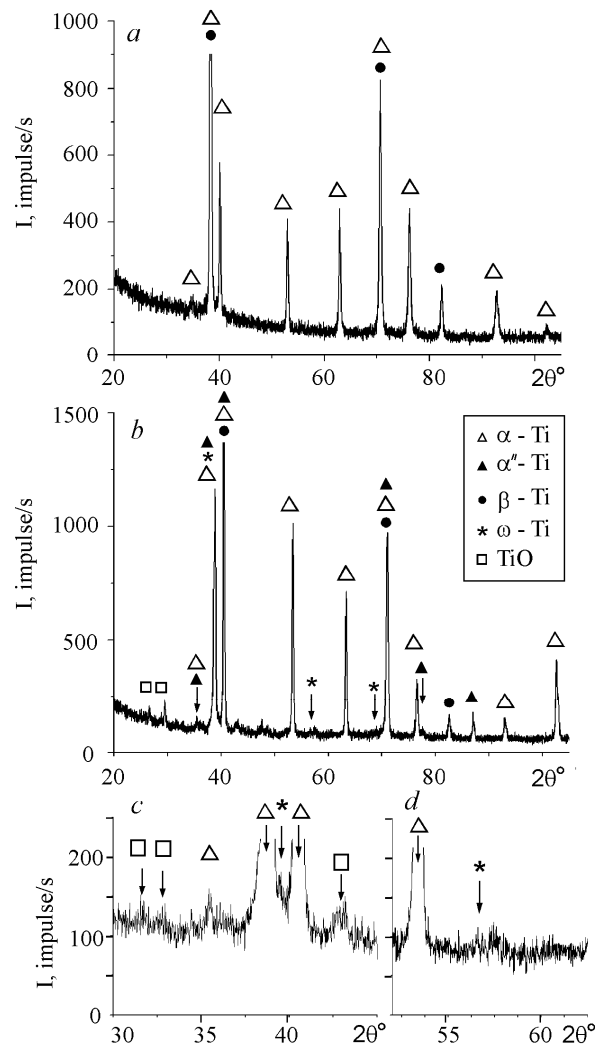


Fig. 2. Diffraction patterns of the alloy Grade2 in the submicrocrystalline state: a is away from the weld; b, c, d is in the weld zone; c, d are fragments of diffraction patterns in the weld zone

This phenomenon consists in a discrete reduction in the transformation temperature when critical cooling rates are reached. Temperature jumps are accompanied by a violation of a monotonous change in the properties and structure of α -titanium caused by different cooling rates [10]. This process can be step-like, and each stage is complex. Thus, formation of stable and metastable phases takes place at the first stage [8]: formation of a stable α -Ti phase and formation of a metastable β -Ti phase. At the second stage, the metastable phase β -Ti can breakdown into a number of metastable phases ω -Ti, α'' -Ti and α_2 -Ti [8, 11].

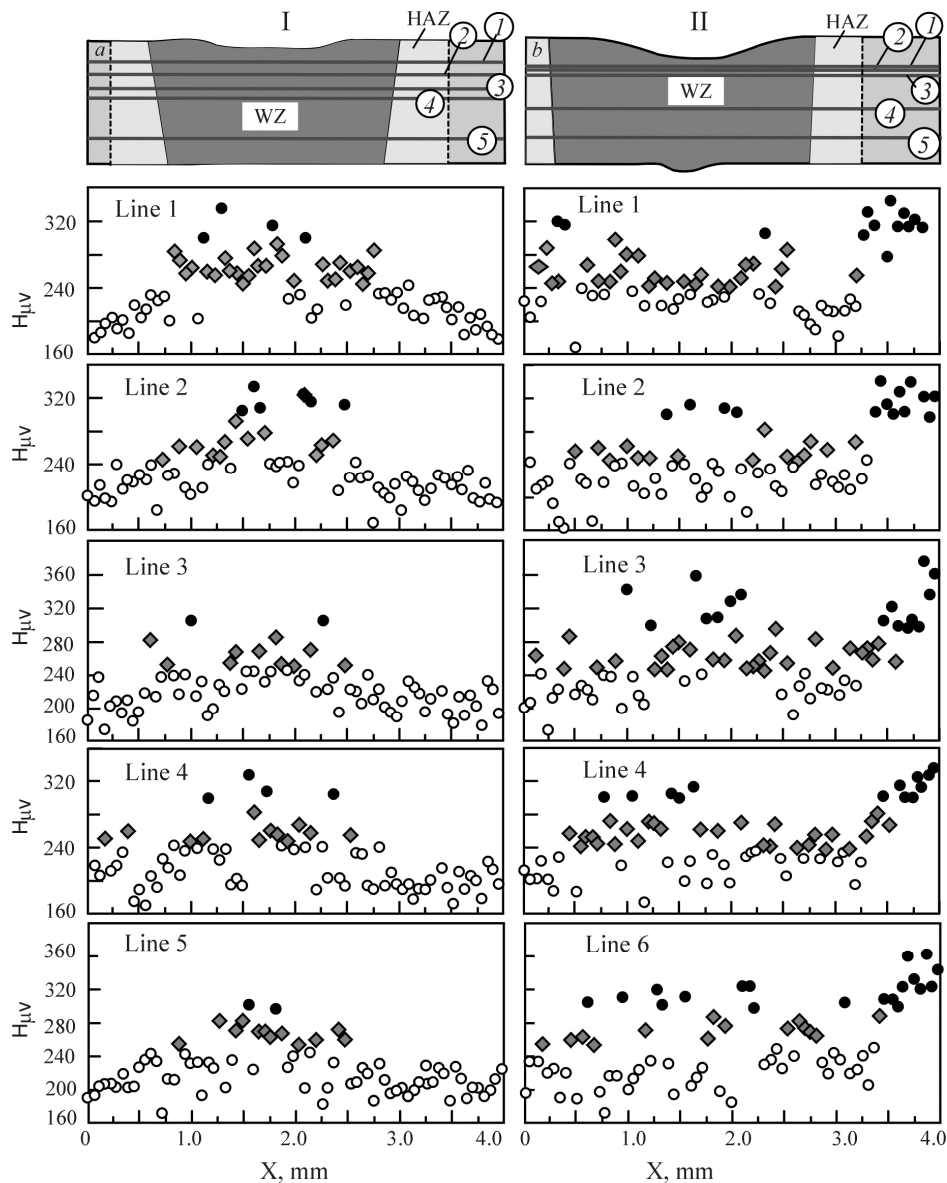


Fig. 3. Schematic representation of heat-affected zones (a, b) and microhardness distribution from the distance along the lines 1, 2, 3, 4, and 5 in the weld zone in the microcrystalline (I) and submicrocrystalline (II) states of the alloy Grade2

Microhardness measurement results

Microhardness measurements in the weld zone and in the heat-affected zone after electron-beam welding on samples of titanium alloys Grade2 in micro- and submicrocrystalline states have been carried out. Fig. 3 shows the microhardness distribution from the distance on the transverse section along five lines on transverse sections of the samples in micro- and submicrocrystalline states of the alloy Grade2. The diagrams show the different character of microhardness distribution depending on the microstructure of the alloy Grade2. Moreover, the obtained microhardness distribution diagrams reflect the weld zones: weld zone (WZ), heat-affected zone (HAZ), and the area away from the weld zone (Fig. 3). It can be seen that in the weld zone and in the HAZ in the investigated alloy, which is in different structural states, there is a considerable scatter of microhardness values. This serves as an indirect estimate of the degree of structural inhomogeneity of the alloy. The obtained data allow to conditionally distinguish three characteristic groups of microhardness values, which are represented by different markers in Fig. 3. The first group of microhardness values is not high and is in the range from 160 to 240 $H_{\mu V}$. These microhardness values are characteristic for the microcrystalline alloy Grade2 [4]. The second group of values $H_{\mu V}$ is intermediate between high and low values, and is in the range of 240 to 290 $H_{\mu V}$. The third group contains high values $H_{\mu V}$

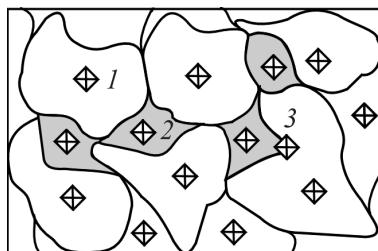


Fig. 4. Diagram illustrating the microhardness measurement in a multiphase material. Numbers indicate the places of indenter “pricks” in different structural components of the alloy: 1 is in a single-phase grain with a low microhardness value; 2 is in the structure region with high microhardness values; 3 is at the interface boundary

(from $H_{\mu V} > 290$). These values are characteristic for the alloy Grade2 in micro- and submicrocrystalline states [12].

Such a spread of microhardness values is caused by the circumstance that the indenter, when pricked on the surface of the sample, falls each time either into different phases or into interfacial boundaries. This phenomenon is illustrated in Fig. 4.

In order to reveal general regularities in the distribution of microhardness $H_{\mu V}$ in the weld zone, the microhardness distribution diagrams $H_{\mu V}=f(x)$ were obtained by summing over the five lines shown in Fig. 3 per one diagram for the alloy Grade2 in micro- and submicrocrystalline states (Fig. 5). It can be seen that distributions of the microhardness $H_{\mu V}$ in the alloy Grade2 in micro- and submicrocrystalline states have a different character. Such a different character of the microhardness distribution $H_{\mu V}$ in the weld zone reflects the manifestation of different structural-phase states in the alloy Grade2 in micro- and submicrocrystalline states. This is confirmed by structural data obtained using the method of X-ray diffraction analysis (Section 3.1).

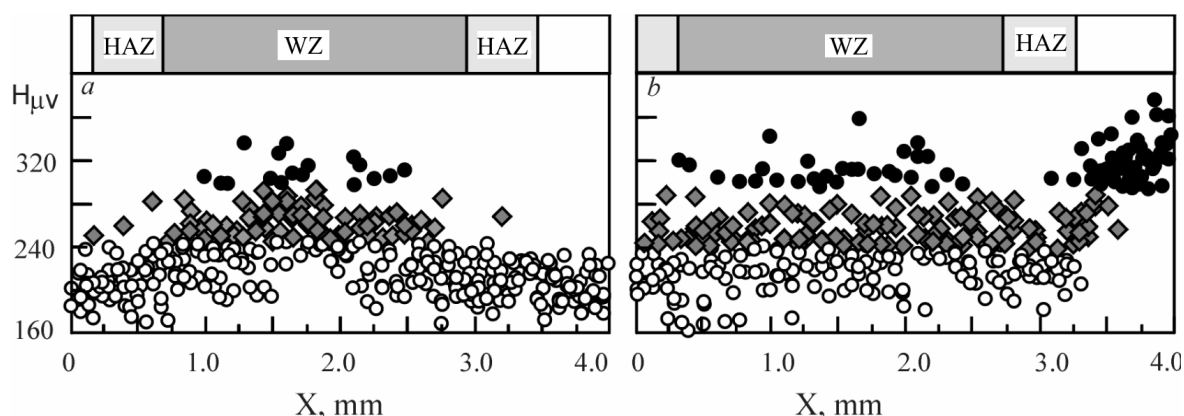


Fig. 5. Distribution of microhardness from the distance summed over the five lines shown in Fig. 3 per one diagram in the weld zone for the alloy Grade2 in micro- (a) and submicrocrystalline (b) states

Conclusion

Thus, the structural-phase state in the weld zone and in the heat-affected zone depends on the state of the samples of the alloy Grade2 when performing electron beam welding: microcrystalline or submicrocrystalline. It has been shown that, in addition to the α' -Ti main phase, the process of formation of metastable phases ω -Ti and α'' -Ti takes place in the submicrocrystalline state in the alloy GRADE2 in the weld zone and in the heat-affected zone.

On the basis of the analysis of data on microhardness distribution in the Grade2 alloy, it is concluded that structural components with higher microhardness values are more frequent in the weld zone in the alloy in the submicrocrystalline state than in the alloy in the microcrystalline state. A crucial point is the fact that, the character of microhardness distribution (as well as the value $H_{\mu V}$) in the HAZ region differs significantly from distributions $H_{\mu V}$ in the weld zone and in the zone away from the weld.

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High-Speed Steel Surface Modification by Electron-Beam Processing

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Keywords: surface modification, microhardness, electron-beam

Abstract. The structure and microhardness of high-speed steels P6M5, P9 and P18 after electron-beam processing were investigated in the work. Electron-beam processing was carried out on the industrial accelerator ELV-4. It was established that electron-beam processing allows to obtain a modified layer on the surface of fast-cutting steels with thickness of 20 µm with high hardness, consisting of fragmented martensite with fine carbide particles. It was determined that after electron beam processing the microhardness of high-speed steels increased to 9.5 GPa. It has been experimentally established that the growth of hardness and wear resistance of high-speed steels after electron-beam processing is the result of the formation of more fragmented martensite and a decrease in the size of carbide particles.

Introduction

At the present time, methods of hardening metals and alloys using concentrated sources of heating-electric arc, ion fluxes, plasma, laser and electron beams, etc., are increasingly used in industrial production [1]. Such sources allows to obtain layers with a high hardness on the surface of the metal, to process surface areas that are inaccessible to hardening by other methods, to localize energy in a narrow zone, providing resource saving [2-4]. Among them, a stream of relativistic electrons is a very effective source of energy that can be used to harden the surface layer of metals and alloys [5-7]. The technology of extravacuum electron-beam processing (irradiation by a stream of relativistic electrons), which is realized with the help of industrial electron accelerators, is characterized by simplicity and volumetric energy input directly into the object of irradiation. Also, this technology meets modern technical, economic, environmental and social requirements. Extravacuum electron-beam processing provides scanning with an electron beam in the air, which allows to increase the area of the processed products and to increase the productivity of the process. The advantages of processing (hardening) by an electron beam include the absence of additional cooling liquids, high efficiency and productivity of the process. Thus, during the processing by an electron beam, significant changes occur in the structural-phase states and in the properties of the material in thin surface layers [8]. Developing in this process of structural rearrangement, structural-phase transformations occur under conditions that are far from thermodynamically equilibrium states, and they allow obtaining modified surface layers with a unique complex of physical and mechanical properties [9].

In connection with the foregoing, the aim of the present work is to study the changes in the structure and microhardness of surface layers of high-speed steels after extra-vacuum electron-beam processing.

Materials and Methods of Research

In accordance with the tasks, the instrumental high-speed steels P6M5, P9 and P18 were chosen as the object of the study. Samples for parallelepiped specimens with dimensions 8x20x20 mm³ were cut from a cutting tool (disc milling cutters) P6M5, P9 and P18 steels with a diamond disk 1 mm thick, which was immersed in a cooling liquid. Extravacuum electron beam processing of samples from high-speed steels was carried out on an industrial electron accelerator ELV-4 [10]. The energy of the beam varied from 1.0 to 1.3 MeV during processing. The blanks were 130 mm from the outlet. Metallographic studies were carried out using the optical microscope ALTAMI-MET-1M.

The morphology of the surface structure was studied with a scanning electron microscope JSM-6390LV, equipped with an attachment of energy-dispersive analysis. Studies of the phase composition of the samples were performed by X-ray diffraction analysis using an X'PertPro diffractometer using $\text{CuK}\alpha$ radiation. The microhardness of the samples was measured by the method of indenting a diamond indenter on a PMT-3M apparatus at a load of 100 g and holding under a load of 10 s.

Results and Discussion

Figure 1 shows the microhardness values of the P9 steels samples surface. P6M5 and P18 are shown before and after electron irradiation. There is a significant increase in the microhardness of the surface of the steel samples after irradiation.

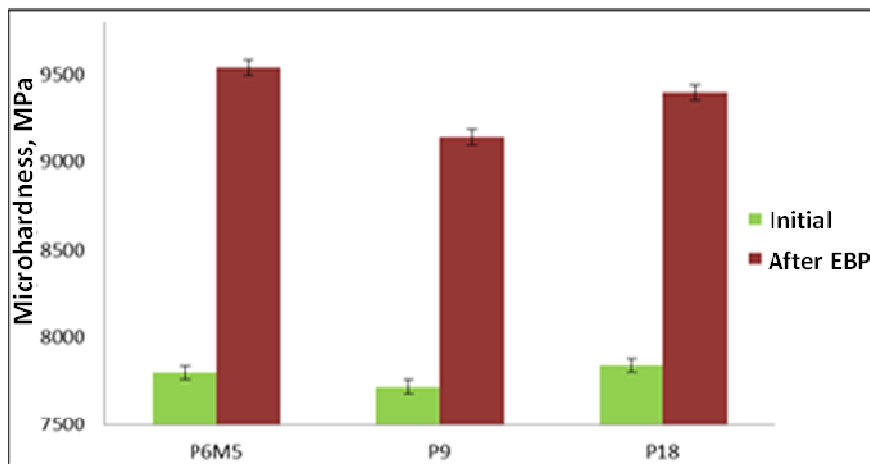


Fig. 1 – Microhardness of P9, P6M5 and P18 steels surface before and after electron beam processing

Structural-phase states of the surface layer of high-speed steels were investigated to determine the structural factors affecting the hardness of high-speed steels.

Figure 2 shows SEM images of the surface of steels P6M5, P9 and P18 before and after electron beam processing. The results of scanning electron microscopy confirm the decrease in the size of carbide particles after electron-beam processing.

Figure 3 shows the microstructure of high-speed steels P6M5, P9 and P18 samples modified surface layer obtained after electron beam processing. It can be seen that after electron-beam processing a hardened layer is formed on the surface, which is a fragmented martensite. At the same time the dark-tanned zone smoothly becomes the basis. The thickness of the modified layer of steels P9, P6M5 and P18 was 20 μm .

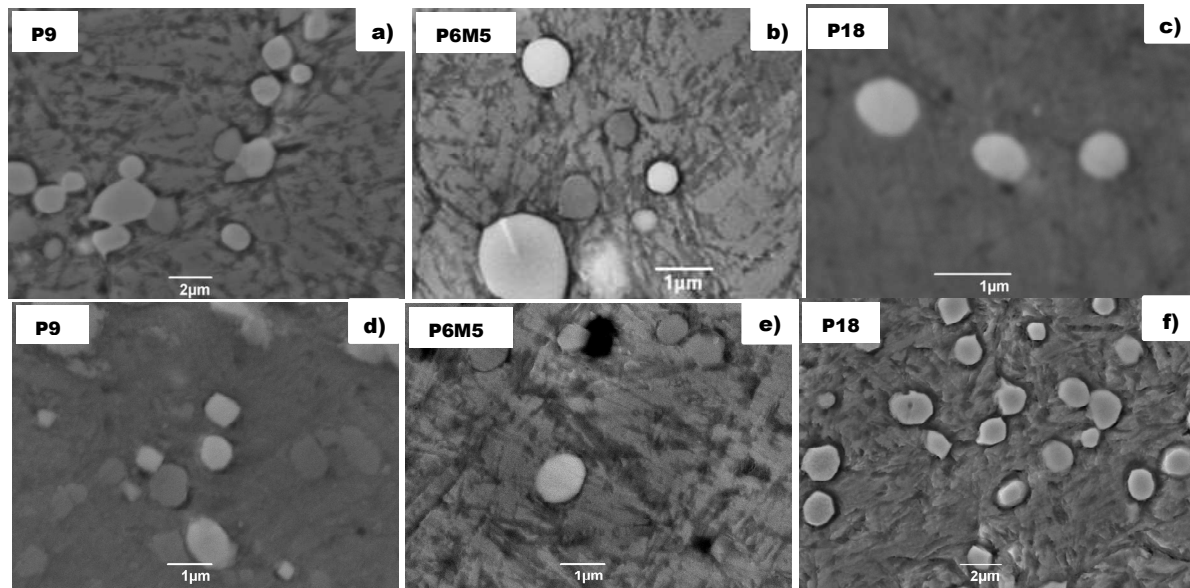


Fig. 2 – Microstructure of high-speed steels surface before (a, b, c) and after (d, e, f) electron beam processing

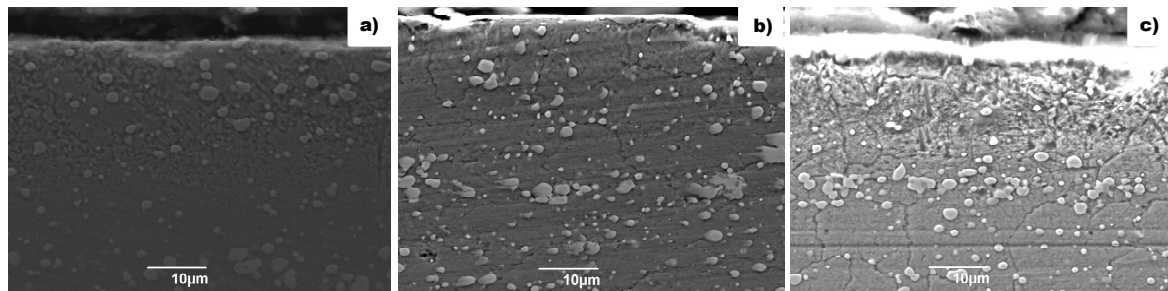


Fig. 3 – Microstructure of a modified layer of P9 (a), P6M5 (b) and P18 (c) high-speed steels after electron-beam processing

An energy-dispersive analysis of the surface of samples of P9, P6M5 and P18 steels before and after electron-beam processing was performed in order to detect changes in the elemental composition of the surface. Figures 4, 5 and 6 show the results of the energy dispersion analysis of the surface of high-speed steels P9, P6M5 and P18 before and after electron-beam processing. Analysis showed that after electron-beam processing, there was no significant change in the elemental composition of the surface of the material. The structure of steels consists of martensite and carbides. It was determined that light spherical carbides were enriched in tungsten and molybdenum, and gray with vanadium. The obtained data confirmed that in the structure of P6M5 and P9 steels there are two types of carbides - light and dark, and in the structure of steel P18 only light. The absence of gray carbides that are enriched in vanadium in the structure of P18 steel is probably due to the fact that the content of vanadium in the composition of P18 steel is small in comparison with the steels P6M5 and P9. In addition, it can be seen from Fig. 6 that the bulk of vanadium is in light carbides in the structure of P18 steel. It should be borne in mind that V, W, Mo and Cr are carbide-forming elements. In other words, the carbides of these metals have a high binding energy and stability [11]. That is why most of the alloying elements are in carbides, and not in solid solution.

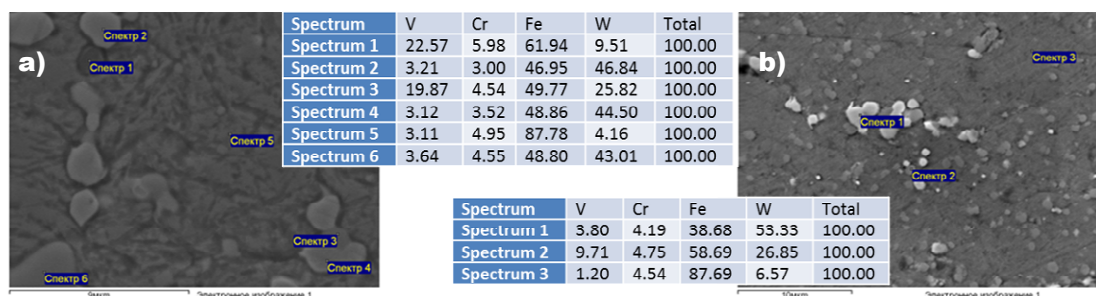


Fig. 4 – Results of X-ray microanalysis of steel P9 before (a) and after (b) electron-beam processing

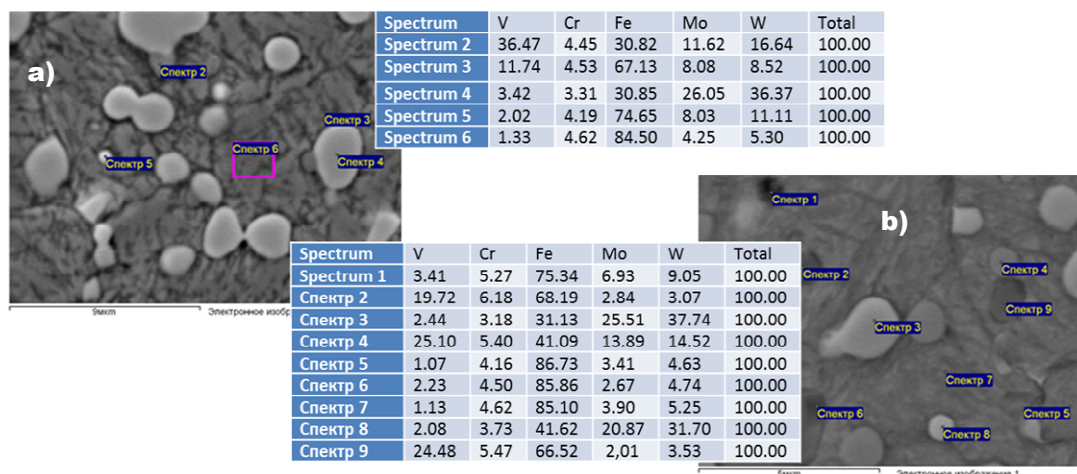


Fig. 5 – Results of X-ray microanalysis of P6M5 steel before (a) and after (b) electron-beam processing

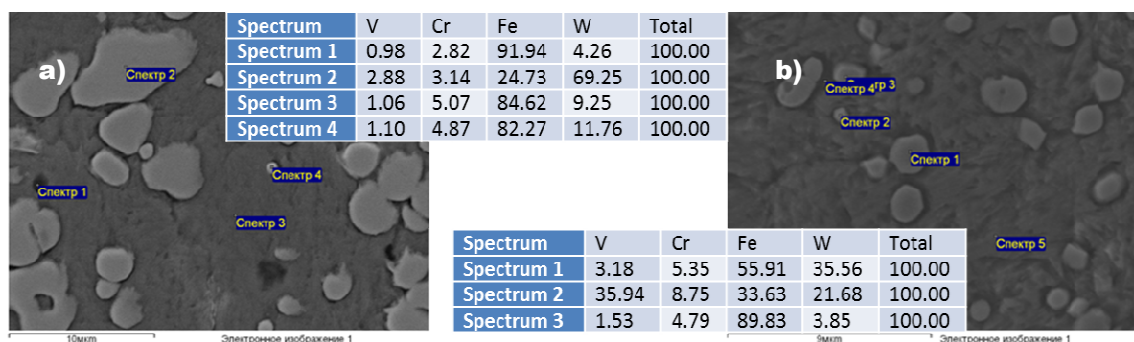
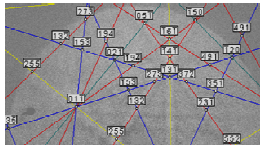
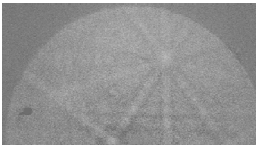


Fig. 6 – Results of X-ray microanalysis of steel P18 before (a) and after (b) electron-beam processing

To reveal the phase composition of the carbides, we determined the crystal structure of the P6M5 steel carbides in the initial state by EBSD analysis using a reflected electron detector on a scanning microscope (system) with electron and focused ion beams (Fig. 7). EBSD analysis has shown that carbides rich in tungsten and molybdenum most optimally combine with the cubic phase of $\text{Fe}_3\text{W}_3\text{C}$, while the carbide rich in vanadium corresponds to the phase of VC.

Thus, the conducted studies showed that a modified layer is formed after electron-beam processing on the surface layer of high-speed steels, which is fragmented martensite. Nevertheless, in order to ascertain the results obtained and to establish the phase composition of the modified layer, we studied phase transformations in surface layers of high-speed steels in electron-beam processing by X-ray diffraction analysis.

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after electron-beam processing is the result of the formation of more fragmented martensite and a decrease in the size of carbide particles.

Thus, the conducted researches have shown perspectives and expediency of application of electron-beam processing for increase of a resource of work of cutting tools.

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Integral Strain Gauges Application for Gearing Reliability Estimation

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Keywords: transmissions, reliability estimation.

Abstract. The paper describes the basic principles and some results of obtaining data by means of integral strain gauges about load-carrying capacity and durability of gear transmissions for new specimens of machines when running in the transmissions in structure of gear box on the stand, or in short-term trials of machines in field conditions.

Introduction

The results of defined serviceability and fatigue endurance of transmissions are used for designer and technological finishing and at attestation trials of transmissions or machines as a whole [1, 2]. The traditional methods of an estimation of reliability of new mechanisms and machines provide long trials of their endurance [3, 4]. To reduce the duration of attestation and certification trials of machines, containing gear transmissions, the accelerated tests of transmissions for finding out the influence of constructive and technology factors on serviceability of transmissions are proposed by the installations of integral strain gauges at the surface of gear-wheel [5]. Various means and tools are invented to monitor the internal state of the elements subjected to fatigue [6, 7, 8, 9, 10]. The most simple and cheapest means are based on the proposed technique of integral strain gauges (ISG), made of a metal foil with special properties, for the total examination and diagnosis of fatigue cumulative damage [11, 12, 13, 14, 15, 16, 17, 18].

Brief Description of Integral Strain Gauges

Integral strain gauges of required configuration are made of metal foil (copper, aluminum, nickel, etc. [3].) obtained by electroplating and subjected to special heat and machine treatment [5, 8, 9, 10, 11, 19, 20]. The size of a source page for manufacturing ISG is 50x50 mm with the thickness of a foil 10 ...30 Micron. Depending on a material of the gauge, structure of electrolyte, parameters of electrolysis and term mechanical processing, and thickness of a foil, the sensitivity of gauges by the number of cyclic strains before appearance of response at given amplitude of cycles varies from several thousands up to several millions cycles [5, 9, 11, 21]. In result of effect of cyclic loading, the structure of a material of the gauge changes and there is the response as "dark spots" on its surface [5, 7, 8, 22]. The moment of appearance of the first "dark spots" and first grains of changed structure of a material of the gauge, and also their density and sizes correlate with number and amplitude of cycles.

Besides described above, some other ways have been developed to estimate the ISG response [5, 11, 17, 19]: by the value of relative square of changed structure, by the change of micro hardness, change of light stream reflected from a surface of the gauge or infra-red radiation.

Implementation of ISG for Gear Transmissions Testing

Firstly, the problems originating from trials for fatigue of gear transmissions on stands should be considered. For such tests, an experimental and calculation method of an individual estimation of serviceability and life prediction of cylindrical gearing [5, 11] (spur, helical, involutes, Novikov) was developed, which is based on the information obtained with ISG after short-term tests of gear box or its part under a load. The full procedure of implementation of the method consists of four main steps of operations.

At the first step, fatigue tests of specimens with a toroidal working part are executed, which are made of the same material and under the same technology as gear wheels. Test results are processed on the basis of the three-parametric Gatts' equation of fatigue curve [5, 11]:

$$N = K \cdot \left[\frac{1}{S - D_i \cdot S_{R0}} - \frac{S_b}{S \cdot (S_b - D_i \cdot S_{R0})} \right], \quad (1)$$

where: N - number of cycles to specimen destruction; D_i - parameter describing damages of a material of a specimen, is a measure of damage reflecting the decrease of endurance limit S_R of the element, which was prestressed, with relevant units: $0 \leq D_i \leq 1$ ($D_i = 1$ is for undamaged material, and $D_i = 0$ when failure occurs depending on the initial value S_{R0} : $D_i = S_R / S_{R0}$); K is for a coefficient; S - is an amplitude of the load cycles; S_b - is an analogues of material strength. Curves created by equation (1) with different D_i are shown in Figure 1. These are basic nonlinear curves in semi logarithmic coordinates, which are used with their modifications for ISG through all the methods. Its main features are that they begin from S_b and end infinitely asymptotically in S_R .

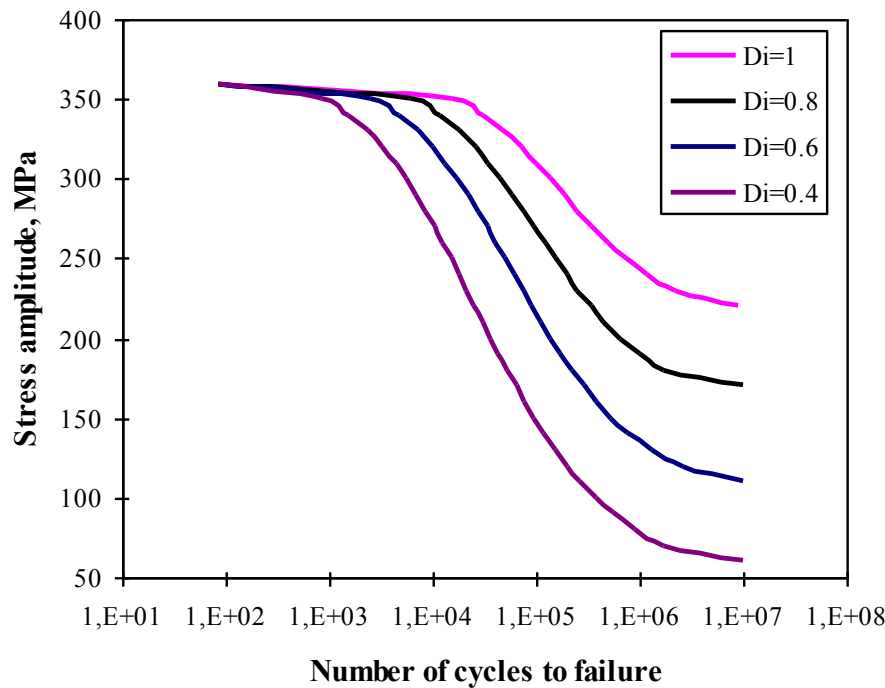


Fig. 1. Fatigue curves corresponding to different degree of damage of specimen material

The final result of the first stage is the equation describing a fatigue curve (with confidence intervals) of a material of the gear wheels of researched transmission:

$$N = K^* \cdot \left[\frac{1}{S - S_{R0}^*} - \frac{S_b^*}{S \cdot (S_b^* - S_{R0}^*)} \right], \quad (2)$$

where values of parameters K^* , S_{R0}^* and S_b^* (at $D_i = 1$) are determined on a base of statistic processing of specimen life tests results.

At the second step, the calibration procedure of ISG is implemented on specimens with a conic working part (Figure 2), and the calibration function with confidence intervals is find out, which describes the relation between the value of stresses fixed by the gauge and appropriate number of load cycles of a calibration specimen. The second step is finished, based on a technique [5, 11, 13], with the definition of parameters of calibration function described with the help of the equation of fatigue curve obtained in the previous step when the damage value, $D_g = D_i$ differs from unit ($D_g < 1$).

As a result of ISG calibration the value $D_g s$ which corresponds to first "dark spots" appearance and also calibrating dependence are determined:

$$N = K^* \cdot \left[\frac{1}{S - D_g \cdot S_{R0}^*} - \frac{S_b^*}{S \cdot (S_b^* - D_g \cdot S_{R0}^*)} \right]. \quad (3)$$

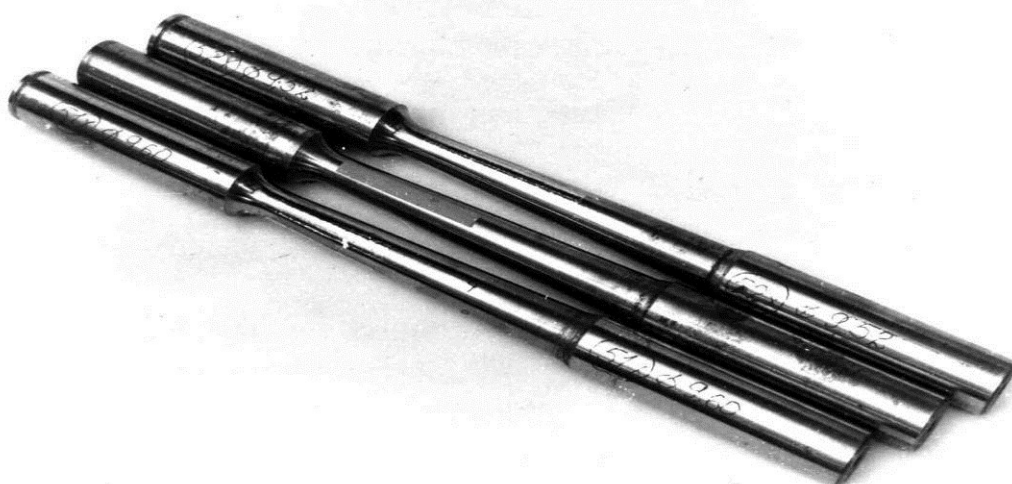
That is the functional link between calibration function (3) for ISG and fatigue curve (2) of a material of gear wheels is installed.

The third step of implementation of the method [11] is based on the theory of similarity, and the transition from the fatigue curve for specimens to fatigue curve for teeth of wheels of researched transmission comes true. The experimental part of this stage provides short-term trials under real loads of real transmission with the errors of manufacture and mounting; ISG are placed in the basis of wheel's teeth. Upon termination of trials on a technique, represented in ref. [5, 11], the maximum stresses in the basis of wheel's teeth are defined by ISG indications at running in of the transmission.

The fourth step of the method is calculation. Using the data obtained after the previous steps the parameters of the equation of bending fatigue strength of teeth are specified, and then reversal problem of loads finding is solved, in result of which the loading, operating in a gearing, is restored. The estimation of number of cycles before loosing the bending endurance is fulfilled on the basis of the known value of stresses (or enumerated for other level of a torque by the specified equation of bending fatigue strength) by the equation of fatigue curve for gear wheels. The forecasting contact endurance is executed on the basic relations represented in GOST 21354-87 and the known value of contact stress derived from the solution of the reverse problem of loading restoration.

The check of a method in the fatigue trials of Novikov's transmission with hard surfaces of teeth has shown the sufficient convergence of predicted results. Failures of gears as teeth breakage occurred in 3.25; 5.13; 6.30; 7.20; 7.70; 12.25mln of load cycles. The forecasted number of cycles (with 95 % of probability) was $N_{\min} = 5.79\text{mln}$, $N_{\text{average}} = 8.6\text{mln}$, $N_{\max} = 15.5\text{mln}$. It should be pointed out that the duration of the transmission trials with ISG application, under a load being necessary for obtaining the information from gauges and implementation of a method, took less than one hour of the stand testing.

When a transmission is tested at a stand the value of torque and the number of load cycles is easily fixed, moreover there is the possibility of numerous interruptions of trials of transmission with the purpose of revealing response on a surface of ISG. At operation trials of mechanisms and machines, the gear transmissions are subjected to the random spectrum of loads, which parameters are known only in rare cases, thus a periodic survey of ISG response, requiring disassembly and assembly of researched transmission, can be impossible at all.



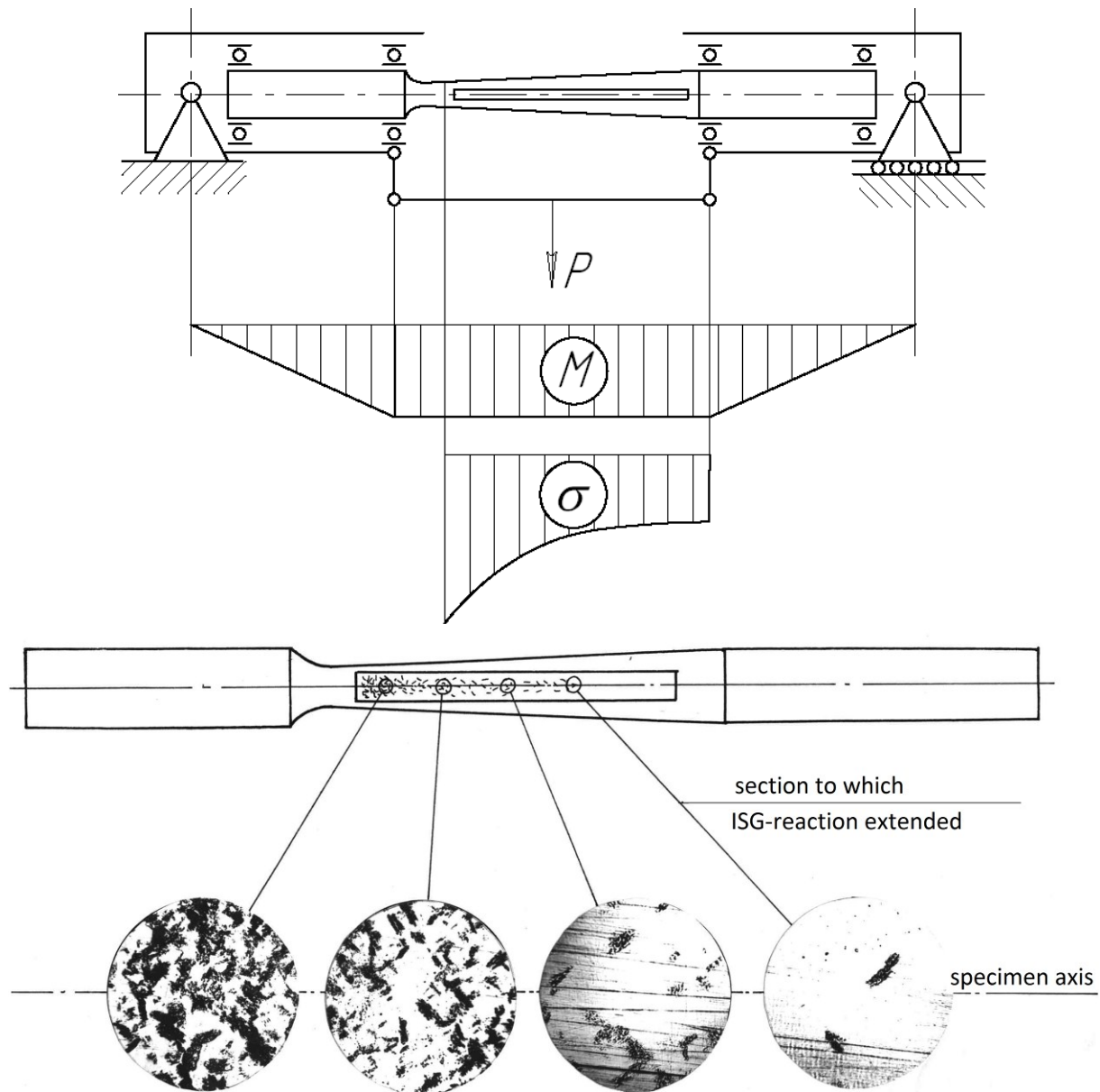


Fig. 2. Calibration of Integral Strain Gauges

The complexity of diagnostics problem of serviceability of gear transmissions in service of machines leads to the necessity of long road trials of machines up to transmissions failure. Therefore, the definition of carrying ability and residual life of transmissions after short operation trials of machines have large scientific and practical significance. The solution of this task has also major commercial value for estimation of quality and reliability of released production. Especially this task is urgent at a stage of putting new specimens of machines in production, when the bank of its failures is default or is rather limited.

The problem stated above was solved during execution of the second and third stages of the program implementation. The first group of the solved problems is devoted to the development of techniques of ISG application ensuring obtaining the necessary estimation of serviceability of gear transmissions in a limiting case for one interruption of trials of transmissions. Several original methods are proposed. The basis of the first method is the use of ISG of various sensitivity [23] to amplitude of cyclic deformations in tests. The second method is for decryption of ISG indications when the results of stress-strain analysis of elements (in the zones where ISG are placed) are obtained with the help of the numerical methods of the fracture mechanics elasticity are attracted [12, 24, 25, 26]. The idea of the third method consists in a new way of registration of ISG response, namely on changing the reflection coefficient of a surface of the gauge in stream of infra-red radiation [5]. This method considerably expands a range of numbers of load cycles, at which ISG

response can be registered, and it makes able to eliminate a role of the subjective factor at an estimation of ISG response. For implementation of this way, a standalone device is manufactured on the base of an optoelectronic converter.

The second group of methods is connected with decryption of the indications of ISG maintained in gears subjected to random loading. The solution of this important task has become possible after realizing wide experimental researches of specimens with applied ISG of various sensitivity [23] on fatigue. The main result may be formulated briefly: the same fatigue cumulative damage of the gears leads to the same ISG response. This result is not universal, some constrains exist, but invariant principle for certain service conditions have been found. The main idea of the method is to describe the fatigue of gear and gauge by the Gatts' fatigue curves describing their fatigue in the invariant terms of damage. The joint consideration of calibration curves for ISG of various sensitivity placed on a researched site of a detail, the stress-strain state of which within the limits of square of gauges is identical, allows us, at known duration of trials of the machine, to restore the equivalent number of load cycles and the level of equivalent stresses.

Conclusion

The check of serviceability of created methods was carried out during experimental researches with the help of integral strain gauges of main transmissions of sporting trucks KAMAZ, participating in rally Paris - Dakar. For the first time in domestic and, we hope, in world practice of trials at absence of data about load conditions of transmissions in service, equivalent mode of loads of transmissions as the equivalent number of cycles and the value of torque have been found and the reason of failure of transmissions installed in field conditions .

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Investigation of Thermal Conditions of the Aluminothermic Welding and their Influence on the Structure and Properties of Metal Rails

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Keywords: welding, hardness, structure, properties.

Abstract. Rail welded joints are integral part of continuous welded rail. However, they often do not have sufficient reliability during the operation. The article is devoted to the assessment of temperature influence effect on the mechanical properties and structure of weld metal and welded rails. The temperature distribution across the rail section in welded zone during the cooling process of aluminothermic rail welding is obtained using simulation by LVMFlow. The results of the study of hardness and structure of metal rail aluminothermic welded joints are given. It is shown that the hardness of rail welded joints increases from 24 HRC to 38 HRC in the fusion zone of the weld metal and rail metal. It is due to the harmful effects of overheating of the metal during the welding process. The hardness is confirmed by microstructural analysis. Microstructural analysis showed the differences in the grains sizes of metal welded zone and heat affected zone. The structure of welded metal is acicular dendritic. Owing to a difference between structures of the welded joint zones the probability of occurrence of cracks on the boundary of fusion weld and metal is increased.

Introduction

The quality level of the welded rail connections largely depends on the accuracy of the welding process parameters. The method of contact rail welding, which has received the widest distribution on the network of Russian railways, allows providing the high quality of rail joints due to precise control of welding modes. However, this method of obtaining welded joints cannot be used for welding rails in the zone of switch points, rail intersections and in curves of small radius. Aluminothermic welding is used in such zones.

To obtain rail joints with aluminothermic welding, the use of complex expensive equipment is not required. The welded joint is formed as a result of molten metal pouring, obtained as a result of the thermite reaction, into the mold, fixed on the connected rails installed with a gap of 25 mm. Removal of the casting mold, excess metal and mechanical processing of the rail head in the welded joint area after welding is performed with the help of portable equipment driven by the autonomous hydraulic unit. Thus, when the rails are welded in an aluminothermic way, there is no need for energy costs, and there is no need for long breaks in train traffic.

However, along with a significant number of advantages, this welding method has a number of important disadvantages. The quality of the welded joints depends on such factors as the quality of thermite material, the proper installation of welding molds, the control of the preheating rail duration, the duration of soaking until the mold is removed, the displacement of the rails in the process of solidification, etc. Failure to comply with at least one of the above conditions during welding can lead to defects in the weld. Since welded joints are an integral part of a seamless rail track and depend on the reliability and safety of train traffic, a lot of work has been devoted to the investigation of these compounds. In the article [1] is noted that the cause of surface defects in aluminothermic welded joints is a violation of the setting mould technology. Investigations aimed at improving mechanical properties of welded joints due to the doping of a thermite powder are given in [2,3]. The articles contain the results of heat treatment influence [4,5] and surface plastic deformation [6,7] on the structure and properties of welded rail joints. In [8,9] you can see the results of simulating the loading process of aluminothermic welded joints of rails.

This work is devoted to the assessment of temperature influence effect on the mechanical properties and structure of weld metal and welded rails. Unlike the contact method for welding rails, in which

the width of the weld is about 6 mm, the aluminothermic welding method has a significant temperature effect on the welded rails because of the large width of the welded joint of 25 mm. Measurement of the metal temperature in the zone of aluminothermic welded rail joints, evaluation of its variation in the welding process and subsequent machining is difficult because the cast metal, having a high temperature (2300 ... 2800 °C) [10] can damage the sensors. In addition, the mold does not allow the sensors to be installed during the crystallization process or to use a non-contact measuring method. It is possible to evaluate the change in the metal temperature in the region of aluminothermic welded joints by computer simulation.

Experimental Part

Simulation of the aluminothermic welded joint crystallization process was carried out with the LVMFlow software product, in which, based on the finite element method, it is possible to determine the temperature value at any point in the model and determine a temperature field in the selected section of the calculated region. In accordance with the existing technological process for the production of aluminothermic welded rail joints, after the termination of the thermite reaction and filling the mold with liquid metal, the mold is first held for 8-10 minutes, and then it is removed and machined.

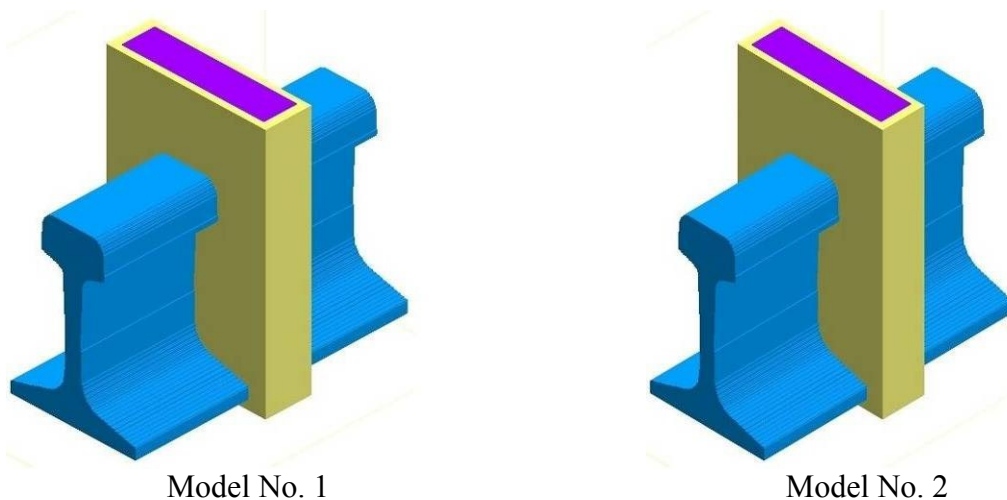


Fig. 1. Three-dimensional models "KOMPAS-3D"

Therefore, for modeling with the help of the program "KOMPAS-3D" two three-dimensional models were created. Model No. 1 consisting of two rails, a welded joint, a mold, gates and a profitable part is designed to simulate the process of cooling the weld metal and the heat affected zone from pouring the metal into the mold until the mold and excess metal are removed (Fig. 1). Model No. 2 consisting of two rails and a welded joint allows measuring the change in the temperature of the metal without a mold (Fig. 1). High accuracy of computations during simulation was provided due to the cells small size of the difference grid equal to 1 mm. Initial installations for modeling the cooling process of a welded joint with a mold (Model No. 1): steel grade – rail steel; the material of the form is quartz sand; the temperature of the filled metal is 2500 °C; the temperature of the rails and mold is 850 °C. Initial data for modeling the cooling of the rail joint without a mold (Model No. 2): steel grade – rail steel; the temperature of the weld metal is 1500 °C, the temperature of the connected rails is 1200 °C.

The temperature of the weld metal and heat affected zones was determined at certain points that were installed on the models in the cross section of the rail and along its length. At the first stage of the simulation, the change in the temperature of the metal of the welded joint was estimated from the moment when the molten metal was poured into the mold until the completion of the mold removal and the excess metal. The second stage consisted in measuring the temperature of the weld metal during the post-welding treatment of the welded joint rail head.

In the study of the thermal conditions influence of the welding process on the structure and mechanical properties of the weld metal and heat affected zones, microstructural studies and measurement of the rail head hardness of the welded joint were carried out. In Fig. 2 is a diagram of the location of the surfaces to be examined and the places where the prints are applied when measuring hardness. Along the length of the welded joint, hardness measurements were made on the rolling surface of the rail (surface A) and at a depth of 20 mm from the rail surface (surface B). In the transverse profile of the rail head, the hardness was measured in the zone of thermal influence at a distance of 20 mm from the fusion boundary (surface B) and in the middle of the weld (surface g).

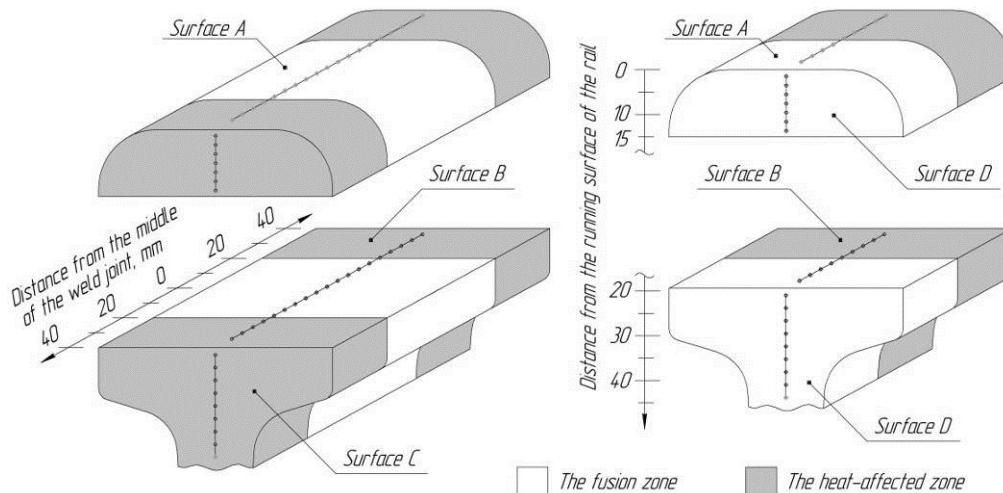


Fig. 2. The location for measuring hardness

Results and Discussion

As a result of computer simulation of the process of cooling the metal of an aluminothermic welded joint, a graph showing the change in temperature at a point established at a depth of 2 ... 3 mm from the rolling surface of the rail head in the middle of the weld (Fig. 3) is obtained.

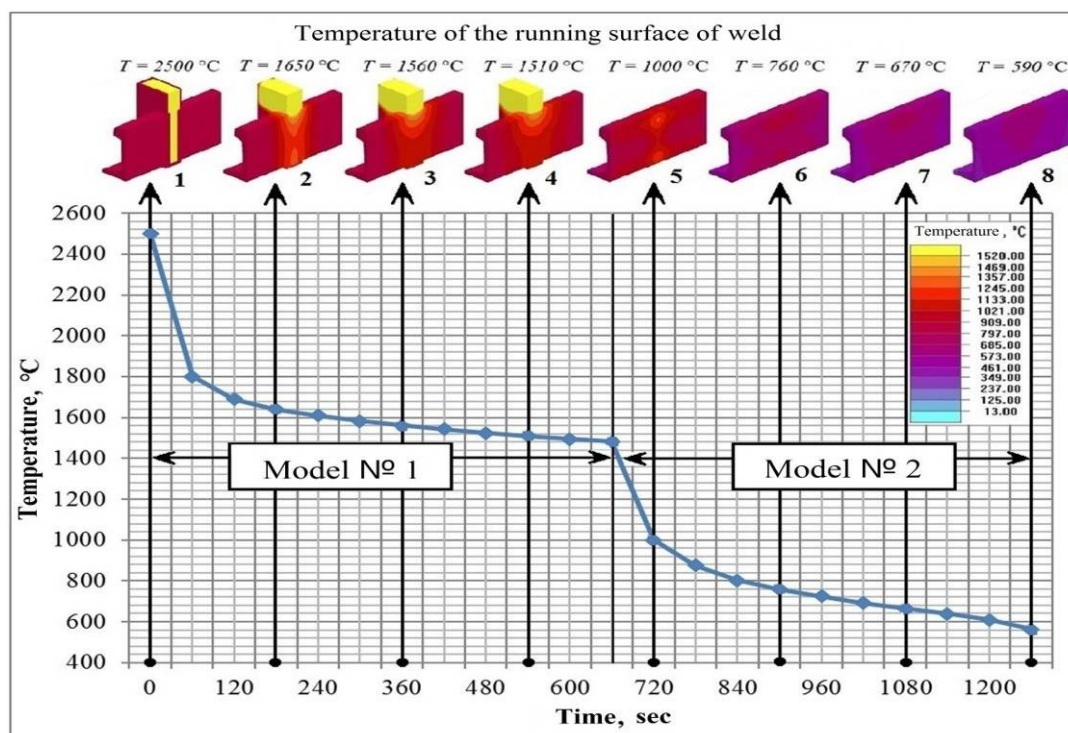


Fig. 3. Temperature distribution across the rail section in welded zone during cooling

After filling the cavity of the mold with a filler metal having a temperature of 2500 °C, an intensive decrease in temperature is observed due to heat dissipation into the welded rails and after 1200 seconds the temperature of the weld metal is 1700 °C. Then the cooling decreases. The metal of the profitable part has the highest temperature, which is confirmed by thermograms (Fig. 3) and prevents cooling of the metal of the rail head in the weld region. After the removal of the profitable part by the casting mold after 660 s, the cooling intensity again increases.

In addition to determining the temperature at a given point, computer simulation allows the graphical representation of the heat distribution across the rail section in the weld zone and in the heat affected zones on the thermograms shown in Fig. 3. Snapshot 1 corresponds to the reference point of modeling the cooling process, the mold is completely filled with liquid metal having temperature 2500°C. Thermograms 2-4 show how the heat from the cast metal is transferred to welded rails, heating them to a temperature of 900 - 1000 °C. At the same time, the temperature of the rail head in the area of the future weld seam remains quite high (1400-1500 °C) due to the heat given off by the metal of the profitable part. After removing the casting mold and removing excess metal, the cooling of the rail head in the weld zone and in the heat affected zones occurs more intensively, which is shown in Fig. 5-8.

As a result of the metallographic analysis it is established that the metal in the region of the aluminothermic welded joint has several different structures (Fig. 4). In the weld zone, the microstructure of the metal is formed during the solidification of the cast metal. Therefore, the metal in this zone has a cast structure of needle-like low-carbon sorbitol with some ferrite along interdendritic interlayers (Fig. 4, a). In the zone of alloying the weld metal with the metal of the rail (Fig. 4, b), a pronounced uneven fusion boundary is visible. The metal of the rail (Figure 4, c) has a coarse-grained structure, which indicates that this area adjacent to the welding zone was overheated and large austenite grains led to the formation of a plate pearlite with a ferrite grid. Overheating of the metal is confirmed by the results of computer simulation, presented on the thermograms 2-4. Areas of overheating in welded joints are dangerous, since the reduced-plasticity and toughness are characteristic of the coarse-grained structure of the metal. Therefore, the probability of cracks in these areas is very high. At a distance of 14 ... 18 mm from the border of a metal rail fusion (Fig. 4, d) is a pearlite structure.

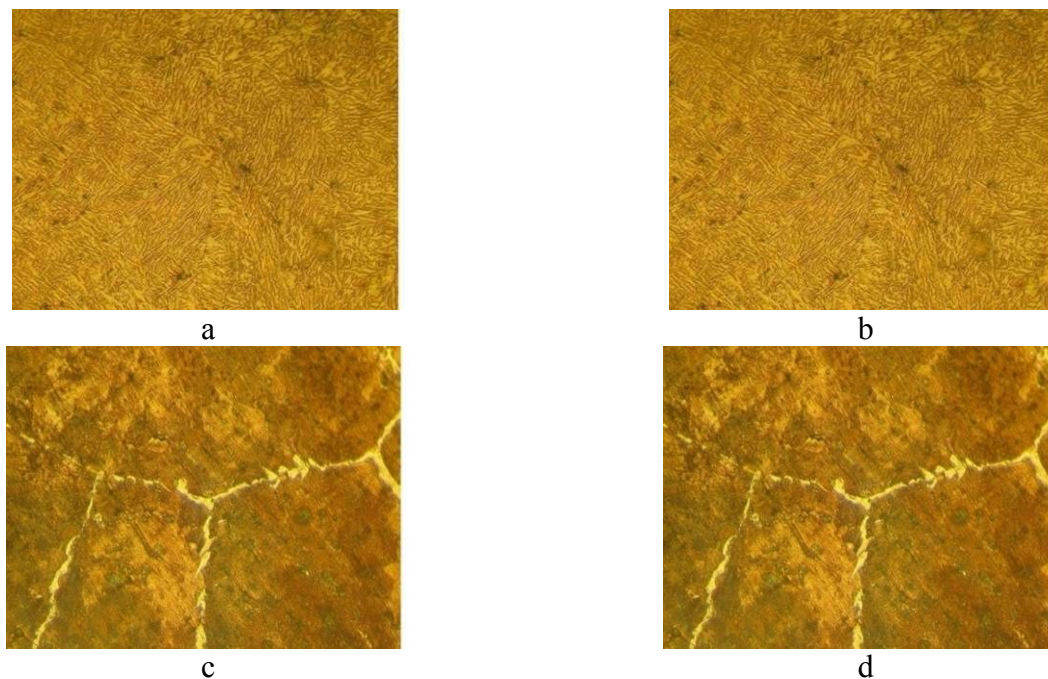


Fig. 4. Microstructure of aluminothermic welded joints. Zone of welded joint (a), zone of alloying the weld metal with the metal of the rail (b), heat-affected zone at a distance of 6...8 mm from the border of a metal rail (c), heat-affected zone at a distance of 14...18 mm from the border of a metal rail (d)

Fig. 5. displays the graphs of hardness variation along the length of the welded joint (surfaces A and B). The graphs are divided into the weld zone and the heat affected zone. The width of the weld was about 50 mm on the surface of the rail and about 40 mm at a depth of 20 mm from it. The boundaries of the transition from the weld zone to the heat affected zone are determined by means of a microscope after etching the surface to be examined.

The hardness of the metal in the welded surface area of the rolling surface of the rail (surface A) is 22...24 HRC. In the heat-affected zone, the hardness is higher, and varies over a relatively wide range from 25 to 36 HRC. The maximum indices of hardness order 36...38 HRC are observed in the zone of thermal influence at a distance of 6...8 mm from the fusion boundary. The increased index of hardness at the beginning of the thermal effect zone is confirmed by micro structural analysis. At a distance of 12...18 mm from the fusion boundary, the indices of hardness vary insignificantly and are about 31...32 HRC. At a depth of 20 mm from the rolling surface (surface B), the change in hardness along the length of the welded joint is similar, the indices of hardness of the weld zone and in the heat affected zone are higher by 1...2 HRC. In the zone of thermal influence, the indices of hardness vary over a wide range and are distributed unevenly along the length of the rail.

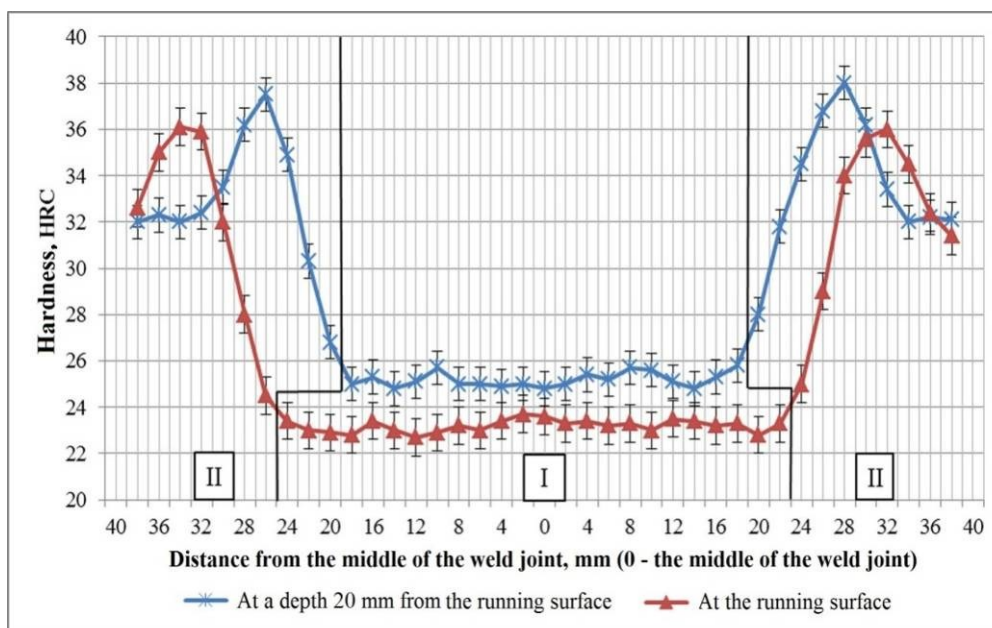


Fig. 5. Hardness of surface A and B. I - The welded zone, II - the heat affected zone

In the zone of thermal influence (surface B), the hardness of the rail metal from the rolling surface to 44 mm along the depth of the rail head varies in a small limit and is about 32...33 HRC. The hardness of the weld metal in depth (surface D) from the surface of the skating to 38 mm lies in the limit of 25 to 28 HRC. At a depth of 38 to 44 mm, the hardness gradually increases to 30 HRC. The increase in hardness is probably associated with a higher cooling rate of the cast metal, since in this region the rail cross-section decreases: the rail head passes into the neck zone.

Conclusion

The results of the simulation clearly show that the temperature in the cross-section of the welded joint rail is uneven: the rail head warms up more than the neck and sole. This is due to the presence of a profitable part of the mold, which prevents the removal of heat. In the process of aluminothermic welding, a significant part of the heat is diverted into the welded rails, which is confirmed by thermograms. As a result of the thermal action on the welded rails near the border of the weld metal and the rail fusion, a region with a coarse-grained structure and an increased hardness of the metal is formed, which is characterized by reduced plasticity and toughness. Reducing plasticity and toughness increases the likelihood of brittle fracture in this area.

Thus, computer modeling with the LVMFlow program provides the possibility of visual analysis of the metal behavior during crystallization of the casting and optimization of the welding process due to changes in the geometry of the mold and the composition of the welding material.

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Modification of Silumin Using the Electroexplosive Method

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Abstract. Surface modification of silumin of the grade AK12 with plasma formed during an electrical explosion of aluminum foil with a sample of yttrium oxide powder placed in the explosion region (the electroexplosive doping method) is carried out. The thermodynamic analysis of the system Al-Si-Y-O is carried out. The phases formed in this system under equilibrium conditions are revealed. Investigations of the elemental and phase composition and the defective substructure are carried out. Mechanical (microhardness) and tribological (wear resistance and friction coefficient) tests of the modified material are performed. The physical mechanisms responsible for multiple enhancement of surface properties of the modified silumin samples are revealed.

Introduction

Silumin is an aluminum-based alloy doped with silicon [1]. The relatively low cost of these alloys and the possibility of widespread use of secondary raw materials during their production allowed silumin to take leading positions in many industries. Silumins are widely used in aircraft engineering, shipbuilding, automotive industry, and domestic use due to their mechanical properties [2]. The most important characteristics of silumin products are microhardness, wear resistance, and fatigue life. To increase these properties, it is needed to reduce the dimensions of all structural components of the alloy (primary silicon crystals, eutectic and aluminum grains, intermetallide inclusions), improve the morphology of silicon crystals (dispersion, conversion of dendrites into equiaxed crystals, spheroidization, uniform distribution over the volume of the alloy), and to carry out doping with interstitial and substitution elements [1-3].

Today, treatment methods based on the use of concentrated energy fluxes (plasma flows, powerful ion and electron beams, laser beams, etc.) are used to implement these structural and phase mechanisms of silumin strengthening [4-6]. Pulsed fusion with simultaneous saturation of surface layers of the material with doping elements followed by high-speed crystallization and formation of strengthening phases, carried out by plasma formed during the electrical explosion of a conductive material (electroexplosive doping, EED), is one of the promising methods for modifying the structure and properties of metals and alloys [7, 8]. Electroexplosive doping of metals and alloys is carried out by a pulsed action of multiphase plasma jets formed during an electrical explosion of conductors. The electric explosion products are a multiphase system comprising both a plasma component and condensed particles of different dispersity [9].

The aim of the paper is the analysis of the regularities in formation of the structure and properties of the surface layer of silumin subjected to electroexplosive doping.

Material and Methods of Research

The alloy AK12 (eutectic silumin, 12.5 % Si, 2.36 % Mg, 0.6 % Cu, 0.35 % Ni, 0.3 % Fe, rest Al (at. %).) was selected as a research material [10]. Electroexplosive doping of the silumin surface was carried out on the installation EVU 60/10 (Siberian State Industrial University, Novokuznetsk). The following mode of doping with plasma formed during an electric explosion of aluminum foil with a sample of Y₂O₃ powder was used: the exposure time of plasma is ~ 100 μs, accelerating voltage is 2.6 kV; the weight of aluminum foil is about 58.9 mg, the weight of Y₂O₃ powder 58.9 mg. The study of the structure of the modified layer of silumin was carried out using methods of metallography, scanning and transmission diffraction electron microscopy. The properties of the surface layer were characterized by the microhardness value and the wear coefficient.

Results and Discussion

The doping of silumin using an electroexplosive method is accompanied by the action onto the material surface of a plasma jet containing elements of a foil and a powder sample, namely, aluminum, yttrium, and oxygen. Thus, a multielement alloy which includes atoms of aluminum, silicon, yttrium, and oxygen is formed in the surface layer of silumin. Below is an analysis of phases formed in the system Al-Si-Y under equilibrium conditions.

System Al-Si. The system Al-Si belongs to a simple eutectic type with little solubility of components in each other in the solid state [11]. The maximum solubility of Si in solid Al is observed at a eutectic temperature of 577 °C and is equal to 1.5 ± 0.1 at. % (at.). The maximum solubility value of Al in Si is insignificant and is equal to 0.016 ± 0.003 at. % at a temperature of 1190 °C.

System Al-Y. In the system Al-Y there are five intermetallic compounds (Al₃Y, Al₂Y, AlY, Al₂Y₃ and AlY₂), which are formed as a result of different reactions (Table 1). Compounds Al₂Y and Al₂Y₃ melt congruently at temperatures of 1485 °C and 1100 °C, respectively. Compounds Al₃Y, AlY, and AlY₂ are formed by peritectic reactions at temperatures of 980 °C, 1130 °C, and 985 °C, respectively. There are two eutectic reactions in the system: $L \leftrightarrow (Al) + Al_3Y$ and $L \leftrightarrow (Y) + AlY_2$ at temperatures of 639 °C and 960 °C, respectively. The solubility of Y in Al is low. The maximum solubility of Y in Al is 0.052% (at) [11].

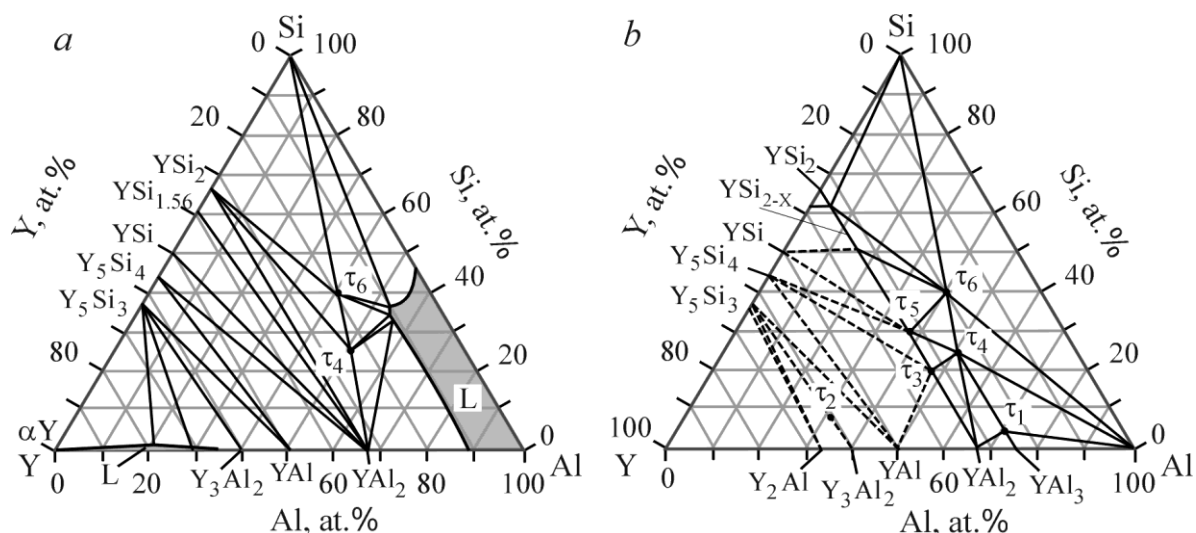


Figure 1. Isothermal sections of the ternary system Al-Si-Y at different temperatures: $T = 1025$ °C (a) and $T = 500$ °C (b) [12, 15, 17, 18].

System Si-Y. In the system Si-Y, the existence of five intermetallic compounds Y₅Si₃, Y₅Si₄, YSi, YSi₂-X (YSi_{1.56}), and YSi₂ has been established (Table 1). Among these compounds, only one (Y₅Si₃) melts congruently at a temperature of 1875 °C. The compound YSi₂ in the system Si-Y is metastable. Two eutectic transformations are observed in the system: $L \leftrightarrow Y_3Si_5 + (Si)$ and $L \leftrightarrow (\alpha Y) + Y_5Si_3$ at temperatures of 1230 °C and 1260 °C, respectively [11].

System Al-Si-Y. In the Al-Si-Y system, among three alloy-forming elements, silicon is classified as a non-metal, since a silicon atom quite easily attaches four more electrons, which characterizes it as a non-metal. Si atoms have four electrons on the outer s- and p- shells – [Ne]3s²3p². The other two alloy-forming elements in this system are metals and have fundamentally different electron configurations: Al – [Ne]3s²3p¹ and Y – [Kr] 5d¹6s². This reflects a key role of electronic factors in the problem of stability and formation of intermetallic compounds based on these elements.

Fig. 1 presents isothermal sections of the ternary system Al-Si-Y at different temperatures [12, 15, 17, 18]. Six ternary compounds have been established in this ternary system (τ_1 , τ_2 , τ_3 , τ_4 , τ_5 , τ_6). The parameters of crystalline structures of three-component compounds are given in Table. 1.

Fig. 2 presents the polythermal section YAl₂-Si in the ternary system Al-Si-Y, obtained on the basis of experimental data [19] and thermodynamic calculations [20]. This polythermal section is quasibinary eutectic. The section YAl₂-Si includes two ternary compounds τ_4 and τ_6 formed as a result of peritectic reactions (Fig. 1, Fig. 2).

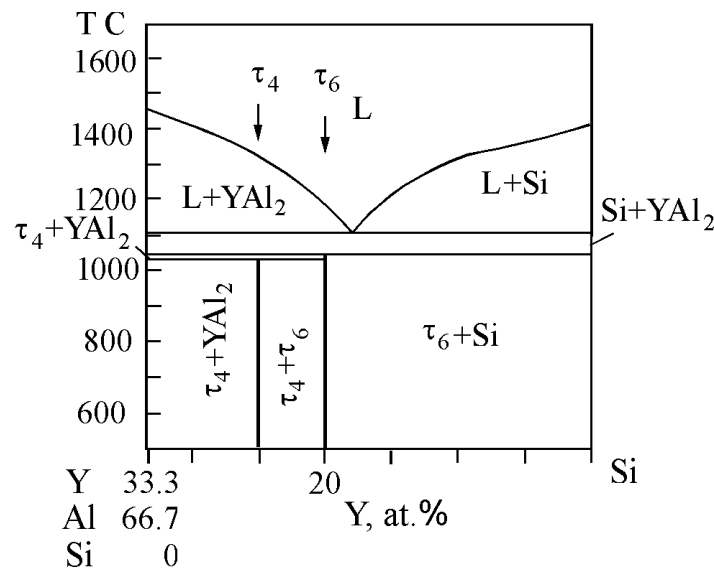


Figure 2. Polythermal section YAl₂-Si [19, 20].

A specific feature of the ternary system Al-Si-Y is the absence of single-phase regions with solid solutions and the presence of three-component intermetallic compounds with certain stoichiometric relationships within the isothermal triangle (Fig. 1, Table). Formation of six three-component compounds τ_1 , τ_2 , τ_3 , τ_4 , τ_5 , τ_6 occurs as a result of peritectic reactions. This leads to the existence of a wide spectrum of two-phase and three-phase regions within the isothermal triangle.

Electroexplosive doping (EED), as a rule, leads to a significant increase in the strength (microhardness) and tribological (wear resistance) properties of the modified material. For treatment of silumin performed in the present work has led to an increase in the microhardness of the modified layer by more than two times, and of the wear resistance by more than 25 times.

Using methods of transmission electron diffraction microscopy, it has been established that the main physical mechanism contributing to an increase in the strength and tribological properties of silumin subjected to electroexplosive doping is the multiple dispersion of the silumin structure [21]. For example, in the initial state, the dimensions of eutectic grains varied within 10 μm , and the dimensions of grains of a solid aluminum-based solution varied within 20 μm (Fig. 3). After electroexplosive doping, the structure of cellular crystallization is formed at the place of eutectic grains (Fig. 4). The average size of cells formed by a solid aluminum-based solution varies within (0.4-0.6) μm . The boundaries of cells include interlayers of silicon, the transverse dimensions of which are (40-50) nm.

Table 1. Compositions and crystalline structures of binary Al-Si, Si-Y and ternary intermetallic compounds Al-Si-Y.

Composition, temperature range, °C	Pearson symbol / Space group	Prototype	Elementary cell parameters, nm			Reference
			a	b	c	
Al-Y						
YAl3(h) <980	hR12	BaPb3	0.6204	–	2.1184	[12]
YAl2 <1485	cF24	Cu2Mg	0.7855(2)	–	–	[12]
YAl <1130	oC8	CrB	0.3884(2)	1.1520(9)	0.4385(2)	[12]
Y3Al2 <1100	tP20	Gd3Al2	0.8232(4)	–	0.7632(8)	[12]
Y2Al <985	oP12	Co3Si	0.6636(6)	0.5086(2)	0.9471(2)	[12]
Y-Si						
Y5Si3 <1850	hP16	Mn5Si4	0.8403	–	0.6303	[13, 14]
Y5Si4 <1840	oP36	Sn5Ge4	0.739	1.452	0.764	[13, 10]
YSi ≤1840	oC8	CrB	4.251	1.0526	0.3826	[9, 14]
YSi2-X <1635	hP3	AlB2	3.843	–	4.143	[14]
βYSi2 1520–1250	tI12	ThSi2	4.04	–	1.342	[14]
αYSi2 <1250	oI12	GdSi2	4.052	3.954	1.3360	[14]
Al-Si-Y						
τ1, Y5Al14Si	hP8	Ni3Sn	0.6247	–	0.4590	[15]
τ2, Y6Al3Si	tI80	Tb6Al3Si	1.163	–	1.512	[12]
τ3, Y33.3Al46.7Si20	oI12	YAl31.4Si0.6	0.4072	0.5812	0.8823	[14-16]
τ4, YAl2Si	mC16	YAl2Si	1.042	0.4070	0.6741	[15, 16]
			β = 101°			
τ5, Y33.3Al36.7Si30	–	–	–	–	–	[15]
τ6, YAl2Si2	hP5	La2SO2	0.4181	–	0.6559	[15]



Figure 3. Electron-microscopic image and the structure of silumin of the grade AK12 before electroexplosive doping (initial state).

The second mechanism for hardening the surface layer of silumin during the electroexplosive doping is the incorporation of yttrium oxide particles Y_2O_3 into the surface layer, i.e. the implementation of the disperse hardening mechanism [22].

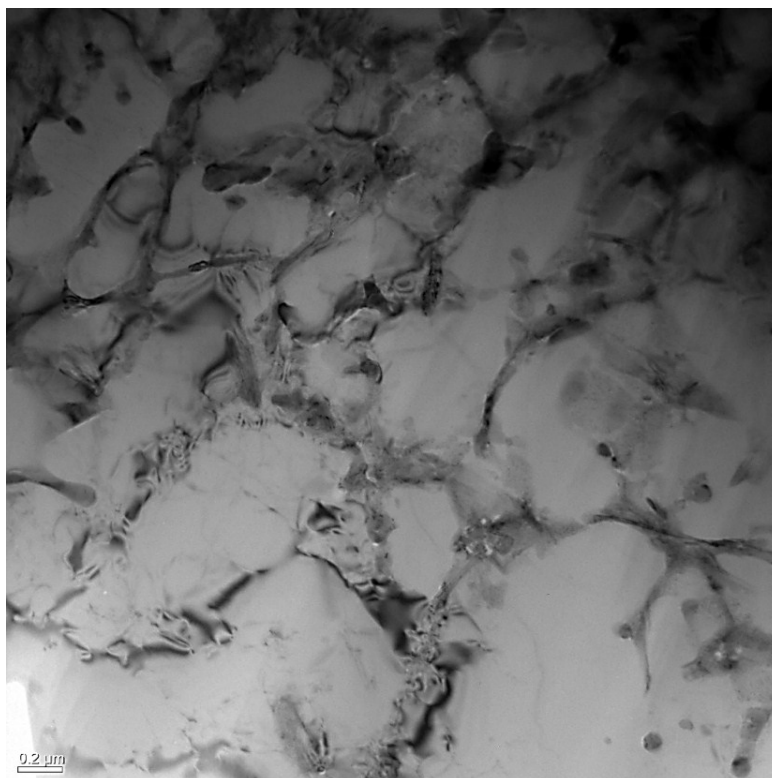


Figure 4. Electron-microscopic image and the structure of silumin of the grade AK12 after electroexplosive doping with yttrium oxide powder.

However, despite obvious advantages of this method, the EED is accompanied by formation of a high-relief, highly porous surface of the modified material (Fig. 5), which makes it very difficult to

use this method as a final finishing of parts and mechanisms. The continuation of this work will be the search for optimal irradiation regimes of the modified EED surface of silumin with an intense pulsed electron beam.

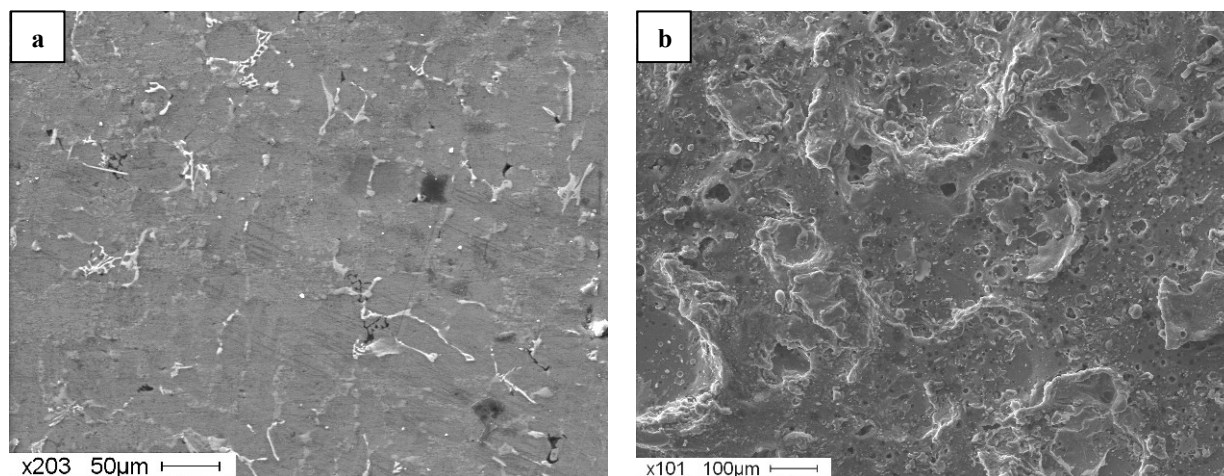


Figure 5. Silumin structure in the initial state (a) and after electroexplosive doping (b).

Conclusion

It has been shown that modification of silumin of the eutectic composition of the grade AK12 with plasma formed during an electric explosion of an aluminum foil with a sample of yttrium oxide powder leads to formation of a modified layer with a thickness of (60-70) μm , the strength and tribological properties of which repeatedly exceed the corresponding properties of silumin in the initial state. Using the methods of scanning and transmission electron diffraction microscopy it has been established that the main mechanisms for increasing the strength and tribological properties of the modified silumin layer are the eutectic dispersion and the incorporation of nanosized yttrium oxide particles into the structure of the surface layer.

Acknowledgement

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On the Thermal and Electrical Characteristics of the Hybrid Plasma-MIG Welding Process

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Keywords: MIG process, Electrical Characteristics, plasma, model.

Abstract. The theoretical analysis of the influence of the plasma discharge temperature with non-consumable electrode on the melting of the electrode wire was carried out. The dynamics of the melting of the electrode wire and dynamics of circuit with consumable electrode arc were been investigation. The estimation of maximum values of the temperature of plasma have been made. Influence of the MIG process on the volt-ampere characteristics of the plasma discharge have been studied. These results were used for carrying out the technological experiments.

1. Introduction

The emergence of new structural materials requires the development of new and improvement of existing welding technologies for the creation permanent connections. Trends in recent decades – the development of hybrid technology [1]. They allow you to combine and amplify the benefits of each of the constituent parts, as well as to eliminate their disadvantages.

Plasma-arc welding technologies began to appear at the end of the last century (e.g., [2, 3]). However, these technologies cannot be called a hybrid because the interaction between the plasma discharge with non-consumable electrode and the arc burning from the end of consumable electrode is extremely weak. Influence of plasma on the melting of the electrode there is also a minimum.

The essence of the process of the hybrid plasma-arc welding (Plasma-MIG welding) [4, 5] is that the consumable electrode arc burns inside the compressed arc non-consumable electrode (Figure 1). Such joint action of the two arcs in one weld pool assists to reduce spattering, increase welding speed, reduce the width of the weld [6, 7, 8, 9, 10]. This causes a decrease mass of the welded structure by reducing the amount of welding wire required for forming the weld in the hybrid plasma-arc welding with consumable electrode [11].

2. Objectives of the Research

The carried out experiments showed the promise of a hybrid Plasma-MIG welding process. The question of carrying out comprehensive studies becomes relevant. Therefore, the purpose of this work was the theoretical analysis of the influence of the plasma discharge temperature with non-consumable electrode on the melting of the electrode wire. In addition, the influence of the MIG process on the electrical characteristics of the plasma discharge should be studied.

The main means was chosen mathematical modeling.

3. The Dynamics of MIG Process with Influence of Plasma Discharge to it

A number of researchers note the great influence of the plasma discharge and the conditions of its burning on the melting of the electrode, the metal transfer and the formation of a weld bead [12, 13, 14, 15, 16].

Both simple [5, 14] and complex [17] models were developed. Complex models are too unwieldy. Technological parameters are difficult to obtain with their help. Simple models give the dependence of the melting rate of the electrode on the plasma discharge current. Nevertheless, experimental studies [18] show that the melting rate of the electrode in the process is almost

independent of the current. Only in the article [3] there is an indication that the melting rate of the electrode is directly determined by the temperature of the plasma discharge. And the temperature of the plasma discharge depends on many technological parameters: current, shielding gas flow, its composition, and so on [12, 19, 20].

We developed the approach proposed in the article [21]. However, the process Plasma-MIG welding cannot be considered as MIG welding with a pre-heated wire because the heating of the electrode wire occurs mainly by plasma radiation, which obeys nonlinear regularities.

3.1. The dynamics of the melting of the electrode wire

Thermal processes in consumable electrode considered under the following assumptions: the phases of heat electrode stick-out and its melting are separated sufficiently and interaction occurs in such a way that the end conditions are heating the initial conditions for melting; electrode stick-out is heated by discharge plasma with non-consumable electrode and by welding current; the length of the melting zone is much smaller than the length of the stick-out; metal transfer is globular. These assumptions are coincide with the assumptions that were made by the author in [21]. Therefore, to construct a mathematical model will go the same way. The difference between the welding processes MIG and plasma-MIG is an electrode stick-out further heated by plasma of discharge with a non-consumable ring electrode (anode).

The electrode wire is located inside a cylindrical plasma discharge on its axis. Increasing of its temperature occurs due to convection heating and radiation heating. Considering that the plasma temperature is $T_p = 5500..15000\text{K}$ is easy to show that the proportion of energy supplied to the wire by convection heating is extremely small. Therefore, when solving the problem of thermal convection heating is not taken into account by us. Heat radiation is governed to the law of Stefan-Boltzmann and the differential equation that describes the change of the temperature T in time t is given by

$$\frac{dT}{dt} = \frac{2\beta \cdot \sigma_{SB}}{\gamma \cdot c \cdot r_w} (T_p^4 - T^4). \quad (1)$$

Here σ_{SB} – Stefan-Boltzmann constant; β , γ , c , r_w – blackness degree, density, specific heat, and the radius of metal electrode wire, respectively.

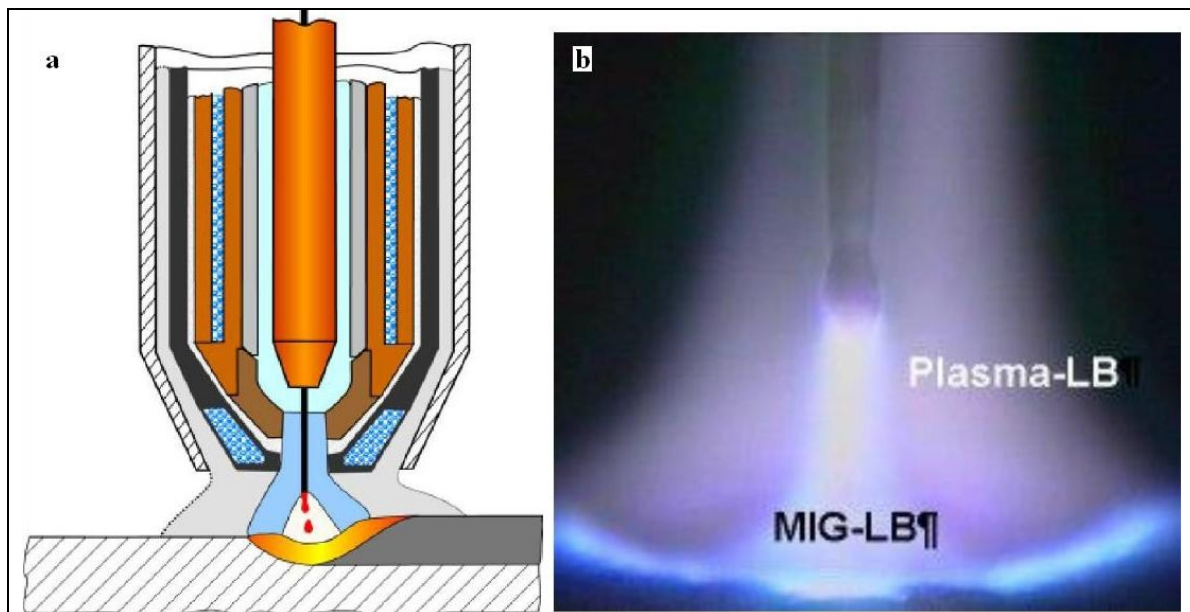


Figure 1. Hybrid welding process Plasma-MIG (a) and photo of interaction between plasma arc and arc with consumable electrode (b).

The solution of the differential equation is an implicit form

$$t = \frac{\beta \cdot \sigma_{SB} \cdot T_p^3}{\gamma \cdot c \cdot r_w} \cdot \left[\operatorname{atanh}\left(\frac{T}{T_p}\right) - \operatorname{atanh}\left(\frac{T_0}{T_p}\right) + \operatorname{atan}\left(\frac{T}{T_p}\right) - \operatorname{atan}\left(\frac{T_0}{T_p}\right) \right]. \quad (2)$$

Here T_0 – initial temperature of the electrode wire.

Using solution (2) for further research is difficult. Simplify it considering the fact that the maximum temperature of the electrode wire melting temperature T_b is much lower plasma temperature T_p . In this case, we get a simple solution

$$T = T_0 + \frac{2\beta \cdot \sigma_{SB} \cdot T_p^4}{\gamma \cdot c \cdot r_w} \cdot t. \quad (3)$$

The difference between the solutions at the melting temperature T_m is 0.2% for steel and 0.02% for aluminum (diameter 1.2mm welding wire). This gave us the basis for further use of the formula (3).

The energy stored in the metal electrode stick-out from passing the welding current and the plasma radiation summed with the energy released in the near electrode (cathode) region is expended on heating and melting metal. Electrode melting speed can be obtained from this condition

$$v_m = \frac{\pi r_w^2 U_k i + \rho i^2 (l_w - l) + 2\pi^2 \cdot \beta \cdot \sigma_{SB} \cdot T_p^4 \cdot r_w^3 (l_p - l)}{\pi^2 \gamma r_w^4 [c(T_m - T_0) + \lambda]}. \quad (4)$$

Here i – welding current (MIG-current process); l – length of the arc; U_k – cathode voltage drop; l_w and l_p – distance from the torch to the workpiece and from the nozzle to the workpiece (see Figure 2); ρ and λ – the resistivity and the latent heat of fusion of the electrode metal.

Obviously, the electrode melting dynamics will be described by a differential equation, whose essence is that the rate of change in electrode length equal to the difference of melting and feeding speed of welding wire velocity v

$$\frac{dl}{dt} = \frac{\pi r_w^2 U_k i + \rho i^2 (l_w - l) + 2\pi^2 \beta \sigma_{SB} T_p^4 r_w^3 (l_p - l)}{\pi^2 \gamma r_w^4 q} - v. \quad (5)$$

Here $q = c(T_m - T_0) + \lambda$.

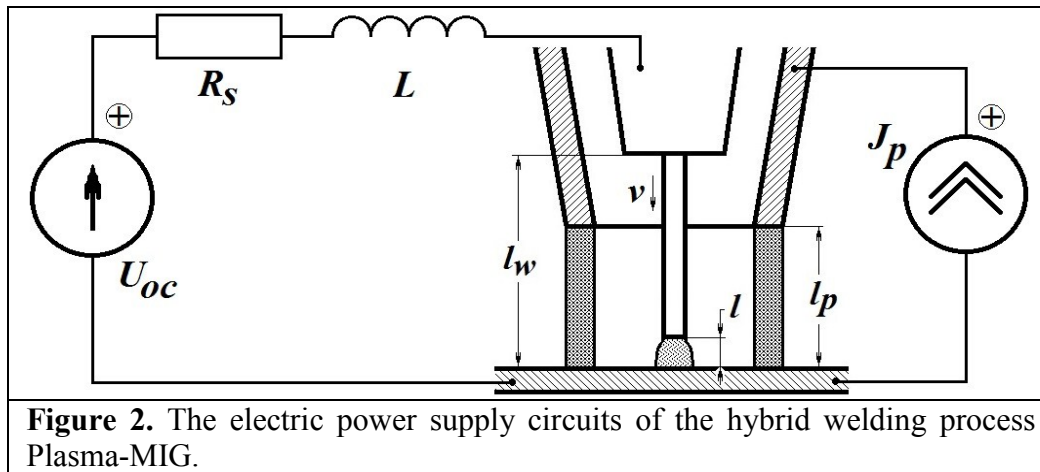


Figure 2. The electric power supply circuits of the hybrid welding process Plasma-MIG.

3.2. The dynamics of circuit with consumable electrode arc

According to Kirchhoff's law, U_{OC} open circuit voltage power supply is equal to the sum of voltage drops on the circuit elements (see Figure 2) – active resistance R_S and the inductance L of the power supply and connecting cables, active resistance electrode stick-out and voltage drop on the arc $E \cdot l$ (the E – electric field intensity in the arc)

$$U_{OC} = R_S i + L \frac{di}{dt} + \rho \frac{l_w - l}{\pi r_w^2} i + El + U_a + U_k. \quad (6)$$

Here U_a – anode voltage drop.

Equations (5) and (6) constitute a system of nonlinear differential equations that describe the dynamics of the system of the "power source - an arc with consumable electrode" in the Plasma-MIG process

$$\begin{cases} \frac{dl}{dt} = \frac{\pi r_w^2 U_k i + \rho i^2 (l_w - l) + 2\pi^2 \beta \sigma_{SB} T_p^4 r_w^3 (l_p - l)}{\pi^2 \gamma r_w^4 q} - v \\ \frac{di}{dt} = \frac{1}{L} \left(U_{OC} - R_S i - \rho \frac{l_w - l}{\pi r_w^2} i - El - U_a - U_k \right) \end{cases} \quad (7)$$

3.3. The static state of "power supply - arc with consumable electrode" system and the condition of its existence

To find the fixed points that define a static state we must equate the right sides of the system (7) to zero. The equations for determining the static value of the arc length l_0 and static value of the welding current i_0 have the form:

$$\begin{aligned} & \rho (r_w E^2 + 2\rho \beta \sigma_{SB} T_p^4) \cdot l_0^3 - \\ & - \rho \left[r_w E (2\bar{U}_{OC} + U_k + El_w) + 4\beta \sigma_{SB} T_p^4 (\pi r_w^2 R_S + \rho l_w) + \right. \\ & \left. + \rho (2\beta \sigma_{SB} T_p^4 l_p - \gamma q v r_w) \right] \cdot l_0^2 + \\ & + \left[r_w U_k E + 2\rho (2\beta \sigma_{SB} T_p^4 l_p - \gamma q v r_w) \right] \cdot (\pi r_w^2 R_S + \rho l_w) + \\ & + \rho r_w \bar{U}_{OC} (\bar{U}_{OC} + U_k + 2El_w) + 2\beta \sigma_{SB} T_p^4 (\pi r_w^2 R_S + \rho l_w)^2 \right] \cdot l_0 - \\ & - \left[r_w U_k \bar{U}_{OC} (\pi r_w^2 R_S + \rho l_w) + \rho r_w \bar{U}_{OC}^2 l_w + \right. \\ & \left. + (2\beta \sigma_{SB} T_p^4 l_p - \gamma q v r_w) (\pi r_w^2 R_S + \rho l_w)^2 \right] = 0; \\ & \rho R_S i_0^3 - \rho (\bar{U}_{OC} + U_k - El_w) i_0^2 + \\ & + \pi r_w \left[2\beta \sigma_{SB} T_p^4 [\pi r_w^2 R_S + \rho (l_w - l)] + r_w (U_k E + \gamma \rho q v) \right] i_0 - \\ & - \pi^2 r_w^3 [2\beta \sigma_{SB} T_p^4 (\bar{U}_{OC} - El_p) + \gamma q v E r_w] = 0. \end{aligned}$$

Here $\bar{U}_{OC} = U_{OC} - U_a - U_k$.

To investigate the properties of these solutions make use of the conditions of existence, the essence of which is as follows: Plasma-MIG process is physically realized when the static welding current is greater than zero

$$i_0 > 0, \quad (8)$$

and static value of the arc length is in the range from zero (contact electrode with workpiece) to a distance from the touch to the workpiece (electrode burning off)

$$0 < l_0 < l_w. \quad (9)$$

Applying these conditions allows defining maximum plasma temperature in which the Plasma-MIG process will carry out

$$T_{p\max} = \left(\frac{\gamma q v E r_w}{2\beta \sigma_{SB} (El_p - U_{OC} + U_a + U_k)} \right)^{1/4}. \quad (10)$$

We have estimated that the temperature (in particular: for steel $T_{p\max} = 7960\text{K}$; for aluminum $T_{p\max} = 7340\text{K}$, process parameters, for which the calculation are given in [4]). This result agrees well with the experimental results of temperature measurements [3].

4. Changing of Electrical Characteristics under the Influence of Shielding Gas and in Interaction with the MIG Process

There is a mutual influence of gas discharges on each other in the hybrid process Plasma-MIG. The effect of a plasma discharge on the melting of an electrode wire was considered above. Melting of the electrode wire in turn affects the electrical parameters of the plasma discharge [17, 22], in particular, its volt-ampere characteristic (VAC).

These features must be taken into account when developing power sources for the Plasma-MIG process [23]. However, the dependence of electrical characteristics on technological parameters of Plasma-MIG process cannot be obtained using existing resistive models [24].

We have improved the arc model developed at the Paton Welding Institute [25, 26] to describe the plasma discharge of Plasma-MIG process.

4.1. Generalized model of welding arc

Let's briefly remind the main statements on which the generalized model of dynamic arc is based.

The plasma arc is phenomenological considered as a thermal macro-object, which is the part of the electric circuit with static VAC $U_{Plasma}(i)$. The energy conservation law for this thermal macro-object looks like.

$$\frac{dQ}{dt} = P - P_{\theta} . \quad (11)$$

This law links three energy parameters of plasma arc column: internal energy Q , input power P and output power P_{θ} . Internal energy of arc column Q is the sum of all plasma energies: heat motion energy, ionization energy, oscillation and rotation energy of molecules, etc. and depends on arc radius and length, i.e. its volume. Under the term of input power P is considered the power, consumed by the plasma arc column from the electric circuit and under the term of output power P_{θ} is considered the power, returned by the plasma arc column to the ambient space by means of different types of thermo-conductivity and radiation.

Application of the arc state current term [25] and introduction of the variable i_{θ} give the opportunity to express the energetic parameters of arc through the electric parameters

$$P = \frac{U_{Plasma}(i_{\theta})}{i_{\theta}} \cdot i_{Plasma}^2; \quad P_{\theta} = U_{Plasma}(i_{\theta}) \cdot i_{\theta}; \quad (12)$$

$$Q = 2\theta \cdot \int_0^{i_{\theta}} U_{Plasma}(i_{\theta}^*) \cdot di_{\theta}^*, \quad (13)$$

(here i_{Plasma} – plasma arc current, θ – arc time constant) and to get the differential equation of the model

$$\theta \frac{di_{\theta}^2}{dt} + i_{\theta}^2 = i_{Plasma}^2 . \quad (14)$$

Equation (14) is supplemented by the formula of connection between plasma arc current and voltage on it

$$u = \frac{U_{Plasma}(i_{\theta})}{i_{\theta}} \cdot i_{Plasma}, \quad (15)$$

which is included into equations, corresponding to the Kirchhoff's laws.

4.2. Generalized model of plasma arc with blow by the shielding gas

In the case of plasma arc with blow by the gas the additional mechanism of energy absorption appears besides the natural convection. This mechanism is based on the forced convection, which power P_v can be taken into account directly

$$P_v = \frac{W_A}{V_A} Q = \frac{v_{ShG}}{l_p} Q . \quad (16)$$

Here V_A is the volume, occupied by the plasma arc column, W_A is the volume of plasma expended per time unit: for longitudinal blown $W_A = \pi \cdot (r_{Aout}^2 - r_{Ain}^2) \cdot v_{ShG}$, v_{ShG} – blow speed by shield gas, r_{Ain} and r_{Aout} are the inner and outer radii of plasma arc column.

The equation of power balance in this case looks like

$$\frac{dQ}{dt} = P - P_\theta - P_v. \quad (17)$$

The equation of arc model with longitudinal blow by shield gas is

$$\theta \frac{di_\theta^2}{dt} = i_{Plasma}^2 - i_\theta^2 \left[1 + \frac{v_{ShG}}{l_p} \cdot \frac{Q}{P_\theta} \right]. \quad (18)$$

Equation (18) is prospective to use for the description of plasma arc dynamics, where the longitudinal blow by the shielding gas is the essential part of the process.

4.3. Generalized model of plasma arc with electrode wire moving inside it

When the electrode wire moves inside a cylindrical plasma discharge, another energy retraction mechanism associated with the heating of the wire appears. It should be taken into account in the energy conservation law

$$\frac{dQ}{dt} = P - P_\theta - P_v - P_w. \quad (17)$$

The electrode wire heating by the radiation of the plasma discharge was considered by us above.

$$P_w = 2\pi r_w \beta \sigma_{SB} T_p^4 (l_p - l_0). \quad (18)$$

The equation of generalized dynamic plasma arc model for Plasma-MIG process

$$\theta \frac{di_\theta^2}{dt} = i_{Plasma}^2 - i_\theta^2 \left(1 + \frac{P_v + P_w}{P_\theta} \right). \quad (19)$$

To obtain the VAC of plasma discharge in the Plasma-MIG process, we use the procedure described in the article [27]. As a result, we obtain a parametric dependence of the voltage on the plasma discharge on its current.

$$U_{PlasmaMIG}(i_\theta) = U(i_\theta) \left(1 + \frac{P_v + P_w}{P_\theta} \right)^{1/2}; \quad (20)$$

$$I_{PlasmaMIG}(i_\theta) = i_\theta \left(1 + \frac{P_v + P_w}{P_\theta} \right)^{1/2}. \quad (21)$$

Values i_θ plays the role of the parameter in these formulas.

The plot presented in Figure 3 demonstrates a tendency to increase the voltage on the plasma discharge with increasing shield gas velocity and its temperature. This is in good agreement with the phenomena that were observed during the technological experiments when comparing Plasma and Plasma-MIG processes [4, 9, 28].

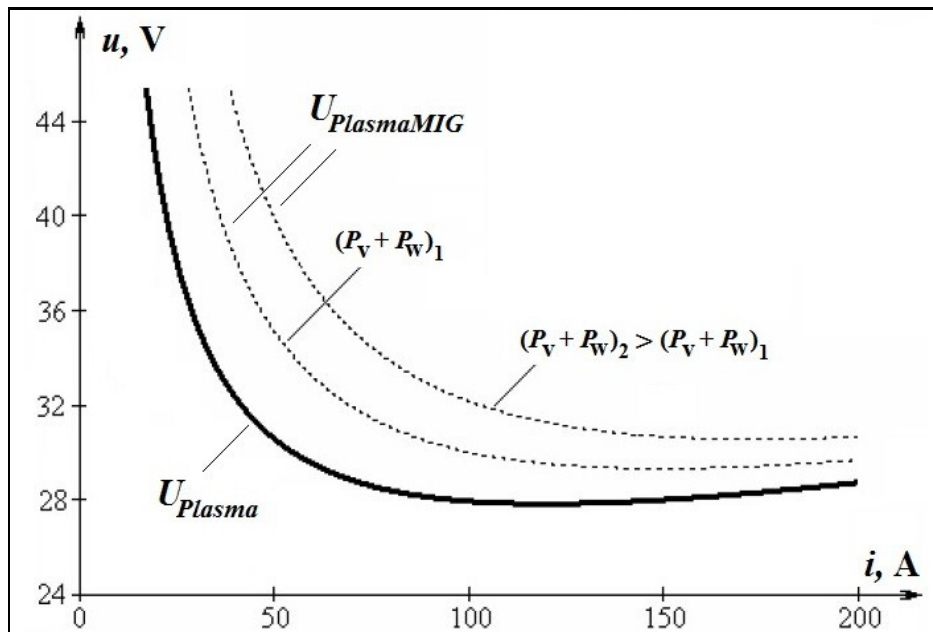


Figure 3. Volt-ampere characteristics of plasma discharge in the Plasma-MIG process (dot lines) in comparison with volt-ampere characteristic of plasma discharge without MIG process (solid line).

5. Equipment and Preliminary Technological Experiment

Technological research of the hybrid Plasma-MIG welding process were carried out using of specialized set of equipment (Figure 4) developed by the authors [5].



Figure 4. Set of equipment for the hybrid Plasma-MIG welding process.

The results of modeling became the basis for choosing of burning mode of plasma discharge.

The calculated process parameters were used in the welding of aluminum alloy and showed good results (see Figure 5). More details were described in [4].

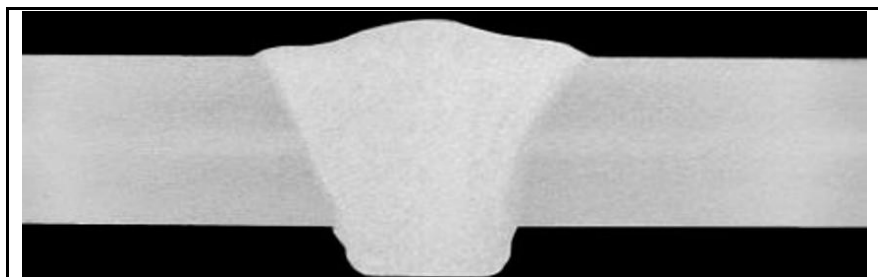


Figure 5. The welded joint of alloy 1561 (8 mm) produced by hybrid welding Plasma-MIG.

Summary

The developed model of heating of the electrode wire in the hybrid welding process Plasma-MIG allow to make evaluation of plasma temperature of the gas discharge with non-consumable electrode.

The obtained differential equations that describe the dynamics of the "power source - an arc with consumable electrode" system, can be used not only to determine the technological parameters of Plasma-MIG process, but also to study condition of its existence and stability.

The application of the law of conservation of energy in the description of a plasma discharge makes as a thermal macro-object it possible to predict the increase in the voltage on it, which must be taken into account when choosing a welding equipment.

Acknowledgements

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Promising Electron-Beam Techniques for Joining Metal Materials in Power Equipment Designs

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Abstract. In paper some design elements are considered, manufacturing of welded joints of which being identified with technological and metallurgical difficulties. The possibility of obtaining welded joints of large thickness parts using EBW with the evaluation of mechanical properties is shown, as well as dissimilar compounds of steel and composite on the basis of tungsten and bronze. The optimal modes of electron beam treatment are picked up and the microstructure of the obtained compounds is investigated confirming their quality.

Introduction

The focused electron beam for manufacturing purposes began to be used from the late 50s of the last century. At the same time, electron-beam techniques have found the widest application in the welding industry due to the possibility of obtaining high-quality welded joints of large thickness in one pass.

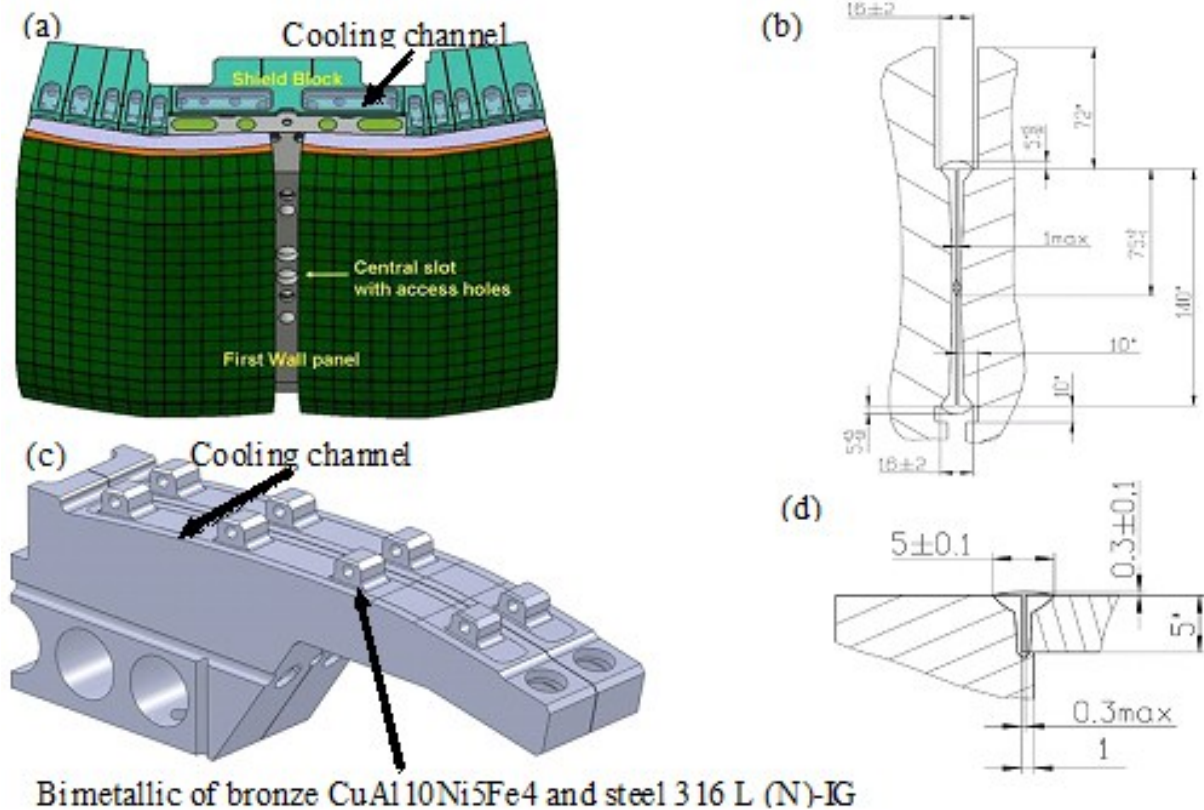
Electron beam welding (EBW) is applied in welding techniques of large thickness parts (high-pressure vessels, turbine rotors, etc.), in creating welded combined structures of dissimilar metals and alloys (diaphragms of steam turbines, bimetallic bandages of electric machine rotors and etc.), in the refractory and chemically active materials structures manufacturing and in joining components in hard-to-get-at places, and in number of other cases [1].

The development of new equipment and the use of new materials are getting more demanding of the welded joints. Meeting of these requirements in turn is possible only with the use of welding methods with wide manufacturing capabilities including the EBW method firstly. Among the manufacturing features of this method distinguishing it from traditional arc welding methods the high energy density in the heating zone, the locality of the thermal action, the smaller welding stresses and deformations can be noted. The other feature of this method expanding its manufacturing capabilities is the possibility of the parameters (power, form of power distribution, speed of movement) and the position of the heating source controlling within wide limits and with high accuracy. For example, modern manufacturing EBW complexes with the maximum beam power allow changing beam current in the range of 1 to 500 mA with practically the same set point error (about 0.1 mA), while the errors of electron gun and the product moving mechanisms positioning being provided at a level of 10-50 µm per 1000 mm of work moving.

One of the power sources of the future is considered to be thermonuclear power being expected to be obtained in the International Thermonuclear Experimental Reactor (ITER). For the thermal protection of the reactor shell and the removal of the plasma thermal power a shield blanket is used consisting of 440 shield modules located on the inner surface of the toroidal vacuum chamber and designed for the following functions: neutron and thermal protection of the vacuum chamber and magnetic field coils against the effects of thermonuclear plasma.

Let's consider some problems needed to be solved when creating the elements of a shield module. The shield module (Fig. 1 (a)) consists of panels of the first wall - a structure facing the plasma; the panels are attached to the supporting structure of the first wall, and the supporting structure of the first wall is connected to the shield block. The basis of the design is the shield blocks consisting of massive parts being joined together by welding. The thickness of the welded joint of the shield block parts is 140 mm while the joint being located in a narrow gap of 72 mm depth (Fig. 1 (b)).

The use of traditional welding methods in this case is impossible for a number of reasons, namely, there is no possibility of making a cutting of the edges as well as of placing a welding torch in a narrow gap [2, 3].



Bimetallic of bronze CuAl10Ni5Fe4 and steel 316 L (N)-IG

Fig. 1. The shield module design scheme of the blanket with the EBW processes. (a) general scheme of the shield module. (b) scheme of welding the blocks together. (c) scheme of the load-bearing structure for attaching the first wall. (d) the shield block cooling channels caps welding scheme.

Another feature of the design is the need for joining parts made of aluminum bronze CuAl10Ni5Fe4 and corrosion-resistant steel 316L(N)-IG (Fig. 1 (c)). These materials are poorly welded by melting due to the low solubility of the components and because of the significant difference in thermophysical properties. The major technique of joining such materials is explosion welding, however this method allows to obtain simple shaped (plate) products only. At the same time when welding these materials, there are no serious obstacles to the liquid phase connection caused by the formation of brittle intermetallic compounds [4, 5].

The supporting structure being the part of the shield module is connected to the shield block. Heat from the structure is diverted by the coolant circulating through internal channels. The channels are sealed by welding the caps (Fig. 1 (d)).

As the special feature of the design being the absence of the direct access to the channels after welding, a cleaning of internal surfaces is not carried out. Accordingly, welding should ensure the cleanliness of the closed space of cooling channels.

All the above design features and requirements for welded joints can be taken into account when using EBW as a single manufacturing method. Such an approach will allow to combine processes carried out to join different parts of the shield module into a unified manufacturing method with its implementation being carried out on a single EBW manufacturing complex.

Another example of challenge while combining materials with sharply differing physical properties is the method of joining a tungsten-copper alloy WCu-80 with chromic zirconium bronze CuCrZr used in the manufacturing of vacuum arc arresters electrical contacts. High-voltage transmission lines are used to transmit electrical power over long distances. Thus, bimetallic elements from the alloy of system W-Cu are used in elements of commutation of high-voltage lines. The W-Cu system

refers to the immiscible components systems being characterized by a large difference in specific masses and melting points and therefore a strong tendency to stratification in the liquid state over a wide range of temperatures and concentrations. Therefore, fusion welding with the mixing of metals is very problematic, because after the crystallization these layers are usually easily separable from each other [6].

Thus, the problems described above when manufacturing elements of modern power equipment can be divided into technological and metallurgical. The technological problems caused by specific requirements to the form and size of the weld or its parts can't be solved with arc types of welding. Metallurgical problems of dissimilar materials joining caused by the specificity of the joined metals interaction in the liquid state and the difference of thermophysical properties requires accurate dosing of the input heat energy to the materials being joined and a high-quality protection of molten metal from atmospheric gases [1,7].

The aim of this work was to study the possibility of using electron beam welding to create special designs of power equipment, including heterogeneous metal materials designs, by an example of the shield module of ITER elements and the vacuum arc arrester contacts.

Research Methodology and the Equipment Used

The researches were carried out using electron-beam complex "AELTC-12" selecting the welding method using the products prototypes. The parts of the products were manufactured from the following materials shown in Table 1.

The method of EBW of the coolant channels caps with the supporting structure body of the first wall and the 5-10 mm thick shield block made of steel 316L(N)-IG. The following requirements are made on the weld joint: the weld must be performed to full depth of the joint and provide the hermiticity under operating loads throughout the entire life of the product; there shall be no lowering of the top seam, penetration into the horizontal gap between the cover and the shield block is necessary to be obtained; there shall be no metal splashing in the inner cavities of the cooling channels. Initially, the welding of the caps was supposed to be conducted according to the lockwise welding scheme (Fig. 2(a)).

Obtaining of lock gap melting is difficult as it is necessary to reduce parameter c to zero and switch to different thickness of edges butt joint. In this case switching to the equal thickness joint is advisable (Fig. 2(b)). However, it did not give the desired results. Therefore, special edge preparation is proposed in order to support the cover, to eliminate the manufacturing gap and splashing in the cooling channels (Fig. 2(c)) [2, 8].

EBW was carried out using "AELTC – 12" installation (Fig. 3). The optimum angle of butt bevel α was experimentally established to be 60° for the conditions presented in Table 2.

The thickness of the shield block welded joint parts is 140 mm. Moreover, the weld joint is located in the manufacturing groove with the width of 8 mm and a depth of 152 mm on one side and respectively with a width of 16 mm and a depth of 72 mm on the other (Fig. 1(b)). Welding of this design is preferable to lead from both sides to align the angular deformation. When using deep penetration EBW, a significant excess weld metal forms on the weld surface being associated with the occurrence of welding deformation and with different density of the core and the cast metal. Thus, in contrast to free surface welding, when welding in narrow gap with deep penetration the molten metal forming the excess weld metal periodically fills the steam-and-gas channel and the appearance of defects, such as a vacuum cavities, becomes possible [1-3]. The optimal double-sided welding conditions are presented in Table 2 [9].

Table 1. Chemical composition of the investigated materials.

Part	Shield block, the supporting structure	Lug	Tungsten-copper tip	Adapter ring, base of the contact
The parts material	316 L(N)-IG	CuAl10Ni5Fe4	WCu-80	CuCrZr
		Percentage, %		
	Fe Basis	Cu Basis	W 80 ± 1	Cu Basis
	C 0.015 – 0.03	Al 9.4	Cu Rest	Cr 0.6-0.8
	Cr 17.0 – 18.0	Fe 4.5		Zr 0.07-0.15
	Ni 12.0 – 12.5	Ni 4.1		
	Mo 2.3 – 2.7	Mn 0.7		
	Mn 1.6 – 2.0	Si 0.1		
	N 0.06 – 0.08	Pb <0.02		

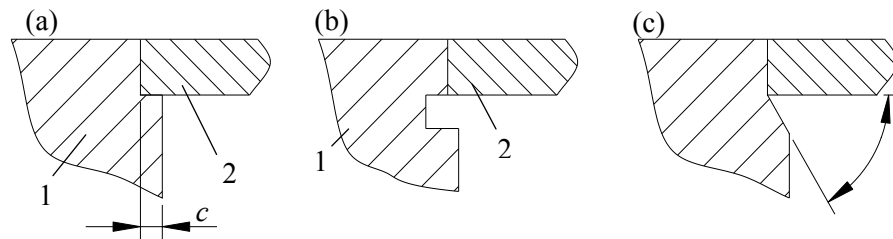


Fig. 2. Coolant channels cap – body weld joint. 1 – housing 2 – cap. (a) lockwise welding scheme. (b) butt welding scheme. (c) bevel butt welding scheme.

Table 2. The recommended welding conditions.

Welded parts	Cap – body	SB modules	CuAl10Ni5Fe4 – steel 316 L(N)-IG		
The thickness of the welded parts, mm	≈12	85	10	20	30
Beam current, mA	100	780	210-230	315-325	510-520
Welding speed, m/h	100	15	30	30	30
Focusing current, mA	963	-	-	-	-
Accelerating voltage, kV	60	60	40	40	40

To eliminate this effect a widening at the base of the groove is proposed to be made, the size of which being determined experimentally based on the size of the excess weld metal when welding at the free surface.

Studies on bronze CuAl10Ni5Fe4 with steel 316 L(N)-IG weldability were carried out on 10, 20 and 30 mm thick samples at different melting degrees of the welded materials. The criteria for assessing the welded joints quality were the presence of defects and the mechanical properties of local areas and a whole weld. A known problem in welding of copper and copper-based alloys with steel is the penetration of liquid copper into steel leading to cracking. This so-called Rebinder effect is associated with the copper on the boundaries of austenitic grains flowing as on the walls of the capillaries and stretching thermal stresses occurring during the welding. To reduce this effect the molten copper with steel contact time is necessary to be minimized. Because of this, optimal parameters of the EBW conditions shown in Table 2 were established [5, 7, 10, 11].

Two ways were used to produce tungsten-copper alloy WCu-80 and chrome - nickel bronze CuCrZr contacts. The first lies in tungsten-copper tip with brass base joining using an adapter ring by soldering, and the second involved the bronze CuCrZr on tungsten-copper tip facing with subsequent welding of the combined workpiece with a bronze base [6].

For both methods implementation, technological samples have been manufactured from the WCu-80 alloy and CuCrZr bronze (Fig. 4).

Soldering, surfacing and welding were carried out on the electron-beam installation with "Languepin-017" power complex with the capacity of 45kW. Before soldering and surfacing, edges of the samples were subjected to machining, electrolytic polishing, degreasing and drying. The parts were collected and fixed in a tool thereafter.

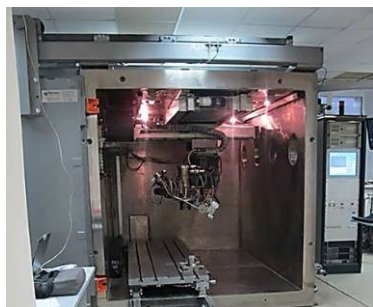


Fig. 3. "AELTC – 12" electron-beam complex.

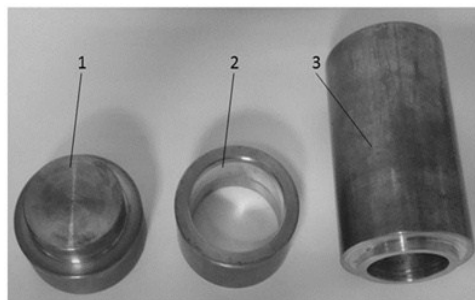


Fig. 4. Workpieces for soldering and surfacing investigation. 1 – tungsten-copper tip. 2 – CuCrZr adapter ring. 3 – CuCrZr base contact.

Soldering was carried out using "STEMET 1108" microcrystalline amorphous solders (Table 3) in a vacuum chamber with a residual pressure of 10^{-4} Torr. The solder was placed between tungsten-copper boss and a chrome-nickel bronze ring and then tungsten-copper boss was heated by defocused beam under the following conditions: accelerating voltage of 30 kV, electron beam current of 80 mA, to a temperature of 30-50 °C above the soldering temperature and kept at this temperature for about 10 seconds. For the convenience of depositing solder into the joint, bronze ring was pre-cut around generating line and, after the solder having been packed, uniformly compressed around the circumference using specialized tooling [6].

Table 3. Chemical composition and properties of "STEMET 1108" solder.

Element	Percentage, %	Tmelting, °C	Tsoldering, °C
Cu	Basis		
Sn	12		
In	7	750 - 870	870 - 950
Ni	2		
P	0.5		

Results and Discussion

Investigations on welded joints of the ITER shield blanket gave the following results. Welding of the coolant channels caps according to the lockwise scheme was found to be technologically difficult (Fig. 5(a)). To simplify the process, an equal thickness joint scheme was proposed. In this case, there is a large amount of splashes in the coolant channel after welding (Fig. 5(b)). The use of special edge preparing during EBW of the coolant channels caps with the body using allowed us to provide a quality weld and to prevent splashing of molten metal in the closed volumes of the channels (Fig. 5(c)). The analysis of the microstructure (Fig. 5 (d)) showed that there are no defects in the weld and all the demands to the weld joint are satisfied when using the proposed welding approaches.

EBW in the narrow gap of the shield block parts (Fig. 6 (a)) allows to obtain quality welded joints. The use of structural widening (Fig. 6(b)) compensates the excess weld metal appearing due to the difference of the base metal and weld metal densities. The application of two-sided EBW (Fig. 6 (c)) allowed to obtain a shield block welded construction without such defects as vacuum cavities with minimal deformation.

The use of the EBW method was established to allow obtaining the bronze CuAl10Ni5Fe4 with steel 316 L(N)-IG defect-free welded joints (Fig. 7(a)). The analysis of the microstructure (Fig.

7(b)) showed that the microdefects in the form of the copper penetration between grains of austenite are also not presented. On the micro-sections steel globular particles can be observed in the weld metal due to the low iron and copper mutual solubility.

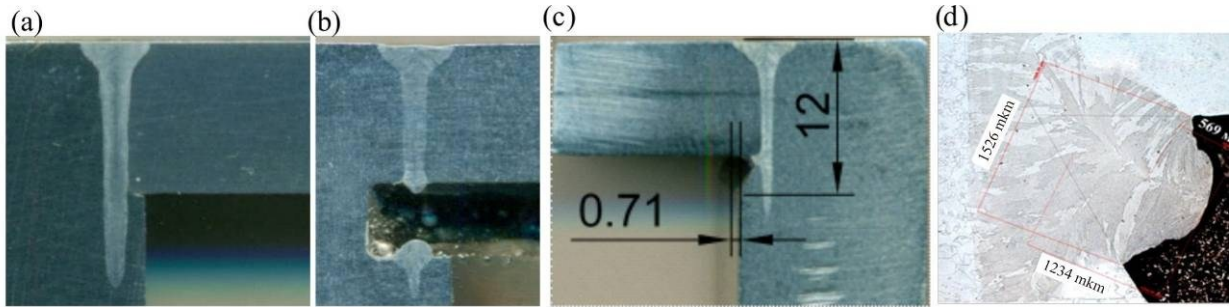


Fig. 5. Weld with different edge preparing. (a) macrosection of a channel cap weld using lockwise scheme with scanning. (b) macrosection of a channel cap weld using butt welding scheme (c) macrosection of a channel cap weld using a special edge preparing. (d) microstructure of the welded joint in the edge preparation zone (X200).

Mechanical tensile tests of welded samples were conducted on Instron 5982 equipment and showed that the destruction is predominantly located in the alloy steel base metal, the main determinant of mechanical properties being the chemical composition of the weld metal: the steel melting degree being 0.2-0.35 and the bronze base metal strength being about 700 MPa, weld strength is at the level of the steel 316 L(N)-IG base metal strength - about 600 MPa. Heat treatment by tempering at 400 °C for 40 min increases the weld metal plasticity [3].

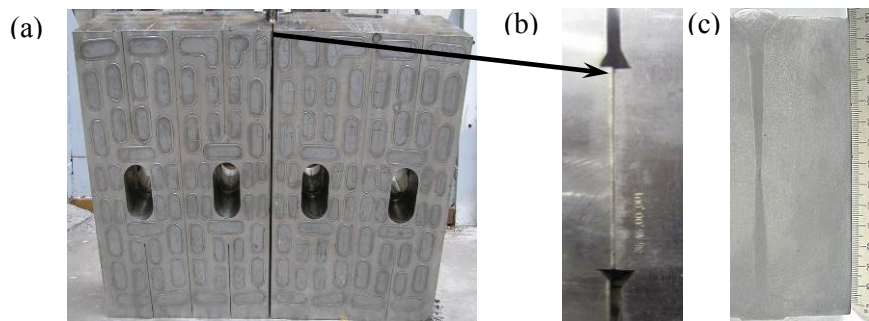


Fig. 6. Welding of the shield block parts (a) external view of the shield block before welding. (b) structural widening in the groove. (c) macrosection of the two-sided welded joint with a thickness of 140 mm.

Researches of bronze with a tungsten-copper composite joining methods showed that joints of satisfactory quality are formed when using “STEMET 1108” solder. However, when the soldered joints being formed, mixing of the molten solder with the composite copper happens on the end surface (Fig. 8(a)). This is due to the fact that heating over the thickness is uneven when using electron beam soldering. The surface layers are heated to higher temperatures than the layers in the depth of the metal. Therefore, when providing the soldering temperature (950°C) on the tungsten-copper tip cylindrical surface, the temperature on its end surface may be greater than the copper melting temperature.

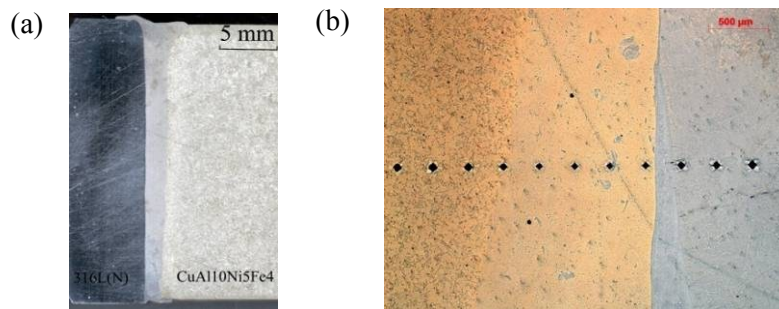


Fig. 7. Welded joint of the 316 L(N) steel with CuAl10Ni5Fe4 bronze sample, (a) macrosection of the steel with bronze welded joint. (b) microstructure (X200) of steel with bronze welded joint with the electron beam being displaced onto the bronze.

Investigation of mechanical properties of the joint obtained showed the tensile strength being determined by mechanical properties of solders and averaged about 160 MPa under normal conditions. As the temperature increase leads to decrease of the mechanical properties of the joint, the maximum operating temperature for this joint shall be around 350°C.

When using another method of joining a tungsten-copper composite with bronze, such as electron beam surfacing, the temperature over the entire surface of the tungsten-copper tip has almost equal values during copper melting and tungsten-copper boss immersion into its melt, thereby letting to obtain even properties and reduce the joint residual stresses. Fig. 8(b) shows a part of bronze with a tungsten-copper composite fusion line, where interface is clearly seen indicating the absence of mixing of this pair of materials, with the bronze having a typical cast structure. Joining of the bronze with a tungsten-copper composite is provided by partial melting of the grains of the copper present in the composite.

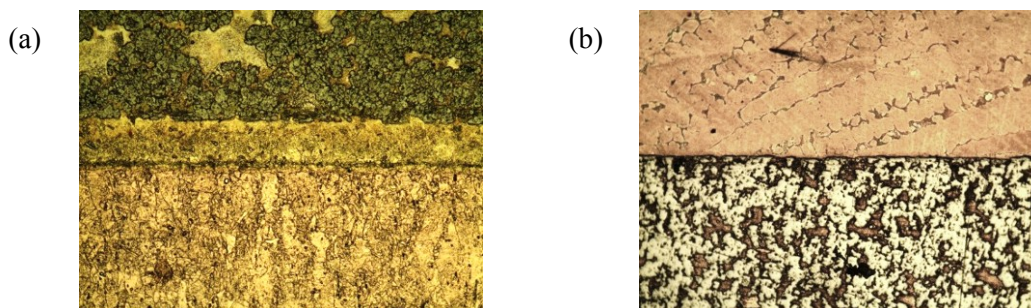


Fig. 8. The photos of microstructures of connections WCu-80 with CuCrZr bronze X200. (a) solder joints (top to bottom: tungsten-copper boss, solder, bronze). (b) the faced layer (from top to bottom: bronze, tungsten copper boss).

Investigation of mechanical properties of joints obtained by surfacing showed the tensile strength to be determined by mechanical properties of copper and to average about 190 - 200 MPa under normal conditions.

Study of properties of welded joints of lined with CuCrZr bronze tip with the contact base made of the same bronze showed that after welding the strength of these connections is not greater than 240 MPa and the plasticity is at the level of 8%. To increase the elastic properties and the tensile strength of CuCrZr bronze welded joints up to 380 MPa a heat treatment is necessary to be conducted: tempering at 990°C with cooling in water and ageing at a temperature of 450°C for 40 min.

However, the increase in plasticity is observed when the aging time is increased up to 3 to 4 hours, with the resistance of welded joints being reduced to 350 MPa during this time.

Conclusions

The use of EBW for joining elements of the ITER shield blanket module will allow to combine single processes in the unified manufacturing method with its implementation being carried out on a EBW universal manufacturing complex. The use of this method for sealing the caps of the shield module cooling system channels will allow to get the smooth reverse roll using the original form of the joint without any additional materials or devices.

When welding CuAl10Ni5Fe4 aluminium bronze with 316 L(N)-IG steel it was found that for obtaining high mechanical properties of welded joints the share of steel in the total amount of welded joints is necessary to be reduced. The optimal value of the steel melting degree is in the range of 0.2-0.35.

Joining of the WCu-80 composite with CuCrZr bronze shall be carried out using an intermediate electron beam surfacing of copper on the composite. The presence of ferrite in the austenitic steel reduces the copper penetration into it as the liquid copper does not wet the ferrite. Allowable depth of copper penetration into steel without affecting the mechanical properties is limited to less than 0.3 – 0.5 mm.

Acknowledgements

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Technology of Powder Coatings Application by Gas-Flame Spraying Using Oxyhydrogen Flame

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Abstract. In the article the equipment, method and the results of researches of strength of gas-flame coatings cohesion with the base are presented. The dependence of adhesive strength on spraying modes of powder material by oxyhydrogen flame is established and described. Recommendations on choosing the gas-flame spraying modes are developed.

Introduction

Successful resolution of issues connected with the loss of metal consumption of construction while increasing its technical characteristics leads to the necessity of composites using. The most progressive process technologies of the materials producing include the methods of protective and hardsurfacing overlay application, a special position is occupied by gas-thermal evaporation processes.

Relevance

It was determined that method of gas-flame spraying of powder materials have been used most intensively in the technology of hardsurfacing and recovering of machine elements. A significant disadvantage of the gas-flame spraying is a separation of an end product of organic fuel combustion while spraying. Application of different fire gases instead of acetylene during the gas-flame processing becomes more and more widespread with every year. The comparative evaluation of methods of detail hardening and restoration by metal coating in terms of the most important performance properties is based on the works of Borisova Ju. S., Vadivasova D. G., Volovika E. L., Kazarceva V. I., Katc N. V., Ljaljakina V. P., Troickogo A. F., Harlamova Ju. A., Hasuj A., Chernoivanova V. I., Shadricheva V. A., et al. As an acetylene substitute a different fire gases are widely used, including hydrogen. Pattern of hydrogen consumption is rapidly changing, especially in the last decade. This effect occurs because the hydrogen is treated as the prospective type of fuel material, universal heating agent and energy storage. Besides, using the hydrogen as a secondary energy carrier has several advantages over hydrocarbon fuel materials. It does not pollute an environment because hydrogen is ecologically pure energy carrier that is still actual. However, widespread implementation the technology is restricted due to absence technological recommendations. That is why development and implementation the gas-flame spraying technology using an oxyhydrogen flame is an actual issue needed to be solved.

Formulation of the Problem

Investigations held by several researchers in exploring the covered detail performance clearly demonstrates that the main problem for destruction is an insufficient connection between covering material and the base [1]. It should be noted that coating separation from the detail surface in the most cases does not correspond the working conditions of the details [2]. By using external force directed tangentially to the coating contact surface, it could be possible to damage the adhesive bond with cutting, shifting or twisting [3]. Studies of the effect of gas-flame spraying modes using oxyhydrogen flame to the strength of cohesion of gas-flame coating to the base is of interest. In order to determine the strength of cohesion it is necessary to shift the coating from the base surface

and to overcome the cohesion among themselves. The value of strength of cohesion depends on the contact area of coating and the base as well as its geometry.

Theory

For the theoretical grounding of oxyhydrogen mixture during gas-flame spraying, comparative evaluation of flame heat qualities of several fire gases was conducted. It was determined that using acetylene flame has the greatest heating efficiency of gas-flame processing, and oxyhydrogen flame has the least efficiency. By examining the obtained results, it can be possible to outline that practically equal heat content of combustion products of acetylene and oxyhydrogen mixtures for normal flame can only be achieved when the ratio of oxygen and hydrogen in the mixture is $\beta_0=0.3\div 0.4$ and the acetylene substitution rate is $\psi=5.2$.

The thermal effect of the constant pressure was determined by thermotechnical evaluation of hydrogen as a fuel gas for gas-flame processing, and thermal effect of the constant volume of chemical reaction of hydrogen combustion (H_2) was obtained. The theoretical amount of oxygen (O_2), necessary for combustion of 1 cubic meter of hydrogen (H_2), was defined.

In addition, interaction of combustion products of an oxyhydrogen flame with material of a base and the evaporated material has been studied.

Experimental Studies

For carrying out the researches the high-alloy structural steel 15HGN2TA TU 14-1-3324-82 was used. The choice of the hard-alloy powder PR-NH17S3R4 TU 14-1-3758-84 of self-fluxing nickel-chromium ferrous alloy with the content of silicone and boron as the material evaporated at the surface of steel 15HGN2TA is proved by the need of the subsequent melting and reproduction of the surface, closest to initial hardness. Fraction of powder is not less than 40 microns and no more than 100 microns.

Drying of powder was carried out in a drying oven SNOL-3,5 at the temperature $t=120\div 150^\circ\text{C}$ during $1\div 1.5$ hour, sifting was made on a shaking sieve 028M.

The surface of a cylindrical sample was subjected to fragmentary thread cutting before a spattering. Thread-cutting machine 1M63, cutter T15K6 GOST18879-86 were used for preparation of the surface.

The technology for a detail surface preparation for a spattering is carried out as follows. The surface of the evaporated detail has previously been processed on the lathe machine, then the «ragged» thread is applied on the processed surface. As surfaces of a detail are cemented on depth at least of $h=0.8$ mm with a high HRC hardness of $58\div 62$, cutting of a «ragged thread» is made on the following modes: thread depth 0,3mm; thread width 0.15 mm; angle 90° ; the shift of the cutting edge from the center of rotation not less than 0,5mm; detail rotating speed of 45 min^{-1} .

The less the time lapse between the preparation of a detail for a spattering and a coating, the higher strength of cohesion of the deposited layer with a detail. The break between the operations of preparation and a spattering should not exceed 3 hours. The detail is heated up in the oven SNOL-3.5 to temperature of 200°C immediately before a spattering.

The powder gas-flame torch Spark-1 of TERMIKA production (figure 1) with the patented nozzle tip of a torch was used for the coating application [4].

The fuel gas was an oxyhydrogen mixture with organic bottled additive of propane-butane no more than 5%. Oxyhydrogen mixture was a conveying gas; in order to obtain the hydrogen water electrolytic generator «Moskva-20» (figure 2) was used.

Oxyhydrogen mixture from the machine is supplied for the oxygen connection of a torch, and propane-butane is brought to the acetylene connection of a torch. Flame capacity can be regulated by the valve «OXYGEN» during the work of the torch (by changing the consumption of oxyhydrogen mixture), and oxidative flame properties are regulated by a valve «ACETYLENE». A purely oxyhydrogen flame is oxidizing; it is possible to obtain the neutral or recovery flame with the addition of carbon-containing gas (propane-butane mixture) in the mixture [4].

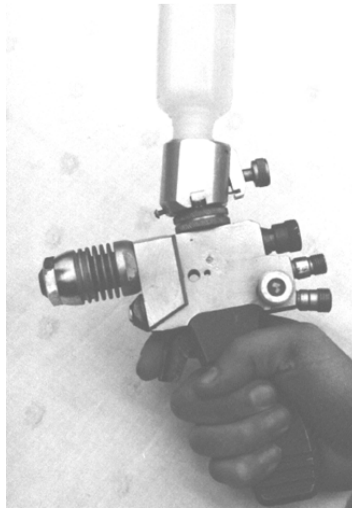


Fig. 1. A gas-flame torch for a powder spraying «Iskra-1»



Fig. 2. Water electrolytic machines «Moskva»

Spraying modes:

- 1) oxygen - pressure $0.4 \div 0.45$ MPa, a combustion $20 \div 23$ l/min;
- 2) acetylene - pressure $0.07 \div 0.1$ MPa, a combustion $10 \div 15$ l/min;
- 3) oxyhydrogen mixture - pressure $0.07 \div 0.1$ MPa, a combustion $16 \div 24$ l/min;
- 4) propane-butane - pressure 0.03 MPa, a combustion $0.8 \div 1.2$ l/min;
- 5) distance of a spraying $120 \div 200$ mm;
- 6) angle of attack 90° , $n=100 \text{ min}^{-1}$;
- 7) fraction of powder $40 \div 100$ micron.

The strongest cohesion of powder material with a base metal (detail) is reached by the melting of the powder after its application to a surface. Melting of the applied coatings was made with a torch «Iskra-2» without supply of powder working on the following gases: bottled oxygen GOST-5583-88 and mixture of bottled propane-butane. An area covered with powder is heated to full melting of all grains of metal in an evaporated layer, until the shining surface is obtained.

Torch «Iskra-2» and the following materials were applied to melt of the applied coatings: bottled oxygen GOST-5583-88 and propane-butane mixture [5].

An adhesive strength at shift is estimated by shift voltage:

$$\tau = \frac{F}{S} = \frac{F}{\pi \cdot b \cdot d}, \quad (1)$$

where F – shearing force, H;

S – area of contact with the base, m^2 ;

d, b – diameter of a sample and the width of coating respectively, m.

One of the most common shearing testing technique is forcing-through of a sample with a coating through a matrix, i.e. strength of cohesion with a base was examined for the shift, according to the technique offered in works [6] by the scheme shown in the figure 3.

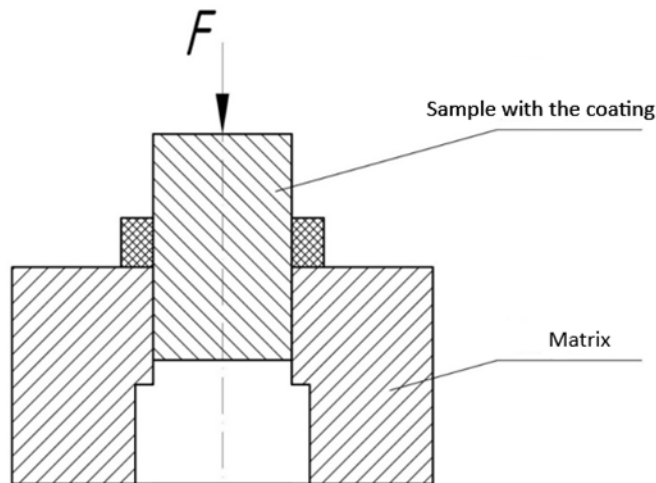


Fig. 3. Scheme of shearing testing of a coating

Examined coating was sprayed on the central part of a surface of a cylindrical sample (base metal). The sample without coating enters a matrix on slip fit; then the sample with a coating is installed in a matrix so that the coating leaned on a face surface of a matrix [7-9].

Samples (figure 4) after the spattering were cooled to room temperature and were tested by forcing-through with the press ZD-10 type production of Germany through the tempered steel ring (figure 5).



Fig. 4. The sample and a matrix for shearing testing technique

The statistical regression model was selected as the mathematical model describing the dependence of change of the adhesive strength of the gas-flame coating to the base. Layered factor experiment of 23 type was designed and conducted to receive an adequate model; the experiment was performed in the local area of time. Value σ of adhesive strength, MPa, has been taken as an output

parameter Y. Based on conducted experiments and literary data, the following major factors influencing the adhesive strength were chosen:

X1 - spraying distance L, mm;

X2 - spraying angle α , degrees;

X3 – fraction of a powder d, microns.

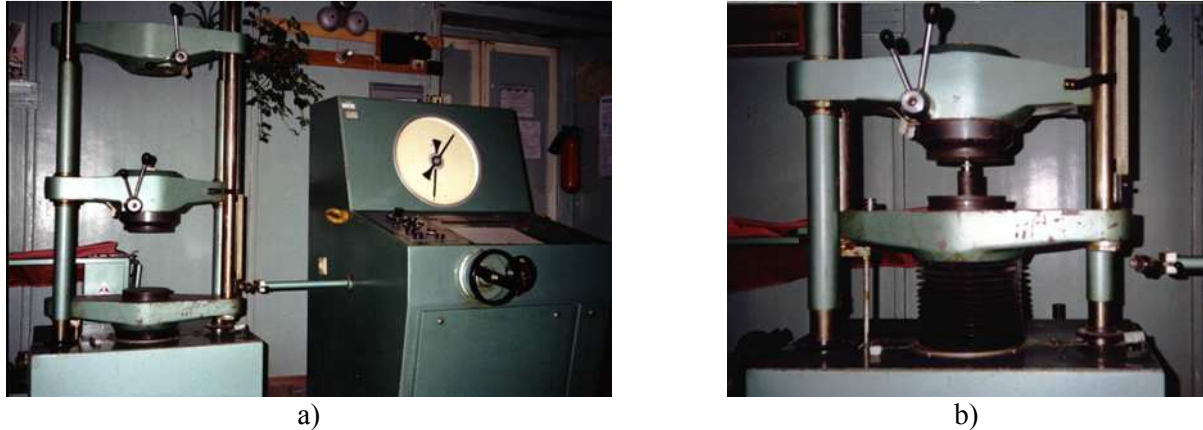


Fig. 5. Definition of adhesive strength (photo): a - a general view; b - press machine

In order to improve adhesive strength of the evaporated and base metal different methods were used [10-12]. The carried-out analysis has demonstrated that it is not possible to achieve good results in all previously offered methods of surface preparation for gas-flame spraying; besides the mechanical connection of sprayed coating with the base, also other types of connection are possible [13-16]. An example is melting of evaporated material with material of the basis forming chemical compound.

Results of a test of definition of the adhesive strength of base detail metal (steel 15HGN2TA GOST 4543-81) and the evaporated powder material (TU 14-1-3758-84 PR-NH17S3R4) are given in figures 6-8. In the studies, the dependence of adhesive strength at an oxyhydrogen flame on the following factors was reproduced:

spraying distance (L, mm);

spraying angle (α , degrees);

fraction of a powder (d, microns).

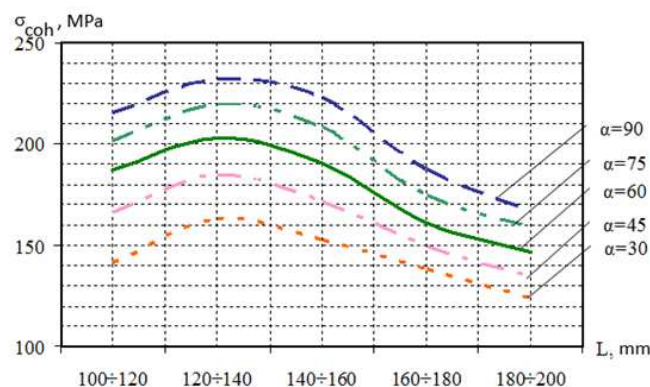


Fig. 6. Impact of L spraying distance on adhesive strength σ_{coh} (powder fraction $d=60\div70$ mm)

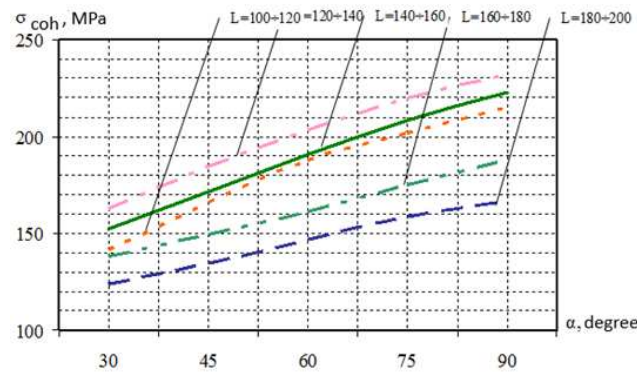


Fig. 7. Impact of α angle of attack on adhesive strength σ_{coh} (powder fraction $d=60\div70$ mm)

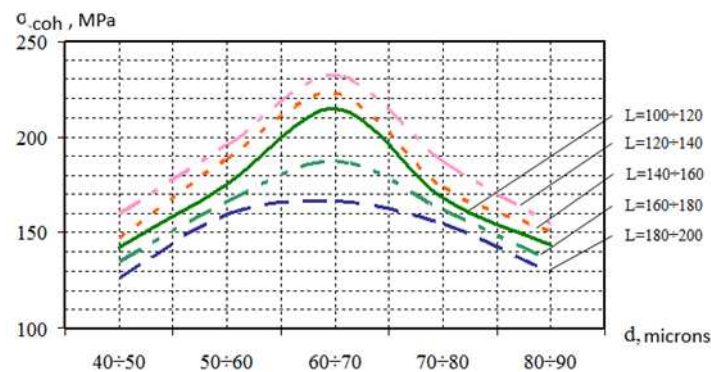


Fig. 8. Impact of d fraction of a powder on adhesive strength σ_{coh} (angle of attack $\alpha=90$ degrees)

The appearance of samples pressed through a matrix after shear test is given in the figure 9.



Fig. 9. Appearance of samples pressed through a matrix after shear test

One of important technology parameters of a spattering is the distance between a detail and the sprayer. Let us consider the influence of a distance of a spattering on the adhesive strength of coatings; then we will analyze physical phenomena that are associated with the ratio. On the base of experimental data, the qualitative chart of dependence of adhesive strength of a coating σ_{coh} on a spattering distance of L (fig. 6) is constructed. The origin of coordinates $L=0$ refers to the place of input of powder into the sprayer, and the zone of heating to initial melting will be characterized by

effective heating of powder and bringing it to the melting. Extension of this zone is close to the length of a kernel of a stream of the gas-flame sprayer. High adhesive strength in a zone of heating to full melting is explained by essential heating of a surface of a substrate by the stream of heated gas. In the case of approach of a substrate to a stream kernel process turns into the known process of facing by powder proceeding with melting of the processed surface.

In case of increasing of a distance in a particle motion zone in a molten condition, there is a gradual falling of adhesive strength, connected with cooling of the evaporated particles [17-19]. However, it has been noticed that the adhesive strength decreases more slowly than the temperature of the particles. Apparently, as a value of L increments, coating increasingly forms from previously overheated particles which keep a fluid phase in the center of the sphere; the hardened particles usually are reflected from a substrate and practically do not participate in the formation of a coating. Therefore, the material utilization coefficient decreases as a distance growing.

The border between a zone of full melting and a zone of particle motion in molten condition can widely vary and depends on the properties of a flame, the applied combustible gas of a stream of the spraying (transporting) gas [19-20].

The final equation of regression takes a form as follows:

$$Y = -501.36 + 1.68x_1 + 0.87x_2 + 16.39x_3 - 0.01x_1^2 - 0.13x_3^2 \quad (2)$$

It follows from the regression equation that independent change of quantitative values of the examined factors influence on the change of adhesive strength; its joint interactions are absent or are less significant.

By studying function on the inflection points by the method of search of optimum value of factors with criterion function $Y = f(X_1, X_2, X_3) \rightarrow \max$, the value $Y = 207.6205235$ was obtained at $X_1 = 118.623649$, $X_2 = 90$ and $X_3 = 64.76237938$ within the examined range with use of Microsoft Excel tools.

The greatest adhesive strength can be obtained at the values of a spattering distance of L (X_1) = 118.6 mm, fraction of powder d (X_3) = 64.8 microns and a spraying angle α , 90 degrees.

Summary

Independent change of quantitative values of the examined factors influences on the change of adhesive strength; its joint interactions are absent or are less significant. The greatest adhesive strength can be obtained with the values of a spattering distance of $L = 115 \div 120$ mm, fraction of powder $d = 60 \div 70$ microns and a spraying angle $\alpha = 90 \pm 5$ degrees.

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Software Application of Material Structure Prediction

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Keywords: Modeling phase transformation, GRID system, software application, digitizing.

Abstract. The developed software application allows the user to model phase transformations under isothermal conditions for a specified steel grade at arbitrary tempering temperature. Architecture of the software application using scheme of the chart of distributed computation of simulation model kinematic parameters for reducing time expenditures during computer experiments. The recognition algorithm of raster images on the isothermal and thermokinetic diagrams has been proposed. The imaging algorithm of the material structure after heat treatment have been developed.

Introduction

When manufacturing various parts, the powder metallurgy techniques are applied. The powder metallurgy components are used at high loads. To obtain required properties of the parts, the heat treatment is used. To predict the properties after the treatment, the thermokinetic and isothermal diagrams are used but they give no idea of material structure.

Computer modelling of phase transformation can give an idea of the resulting material structure. The present simulation models have a number of drawbacks as they are of raster-type. Therefore, at increase in the amount of processing information their resources usage grows exponentially. Besides, they prevent from viewing a complete picture of the material structure.

The Main Body

To simulate a material structure, a software application has been developed. It is a system of the full-range engineering research [1]. The system allows a user to perform a computer modelling of structural transformations in steel under isothermal tempering conditions and continuous cooling, plot isothermal and thermokinetic diagrams, calculate material hardenability, compute main stereometric parameters of the process under consideration according to the results of modelling, save, edit, and view the experimental results.

For all modules to interact and for required functional capabilities to be implemented the software application is based on three-tier architecture (Figure 1).

The terminal is a front-end component. It enables a user to authorize, enter required data, start computing, and view the results. This component is presented as a Windows application connected with the application server being on the FTP server logging on to which takes place by Internet access. The terminal and application server communicate by sending a XML request for performing appropriate calculations via HTTP. After completing calculations on the server side, these results are sent to a client and they are mapped in application interfaces.

To arrange data exchange between a client application and application server, Apache HTTP server is used. The main advantages of Apache are its reliability and configuration flexibility. It allows connecting external modules for data provisioning, using DBMS, modifying error messages, flexible controlling resources access [2].

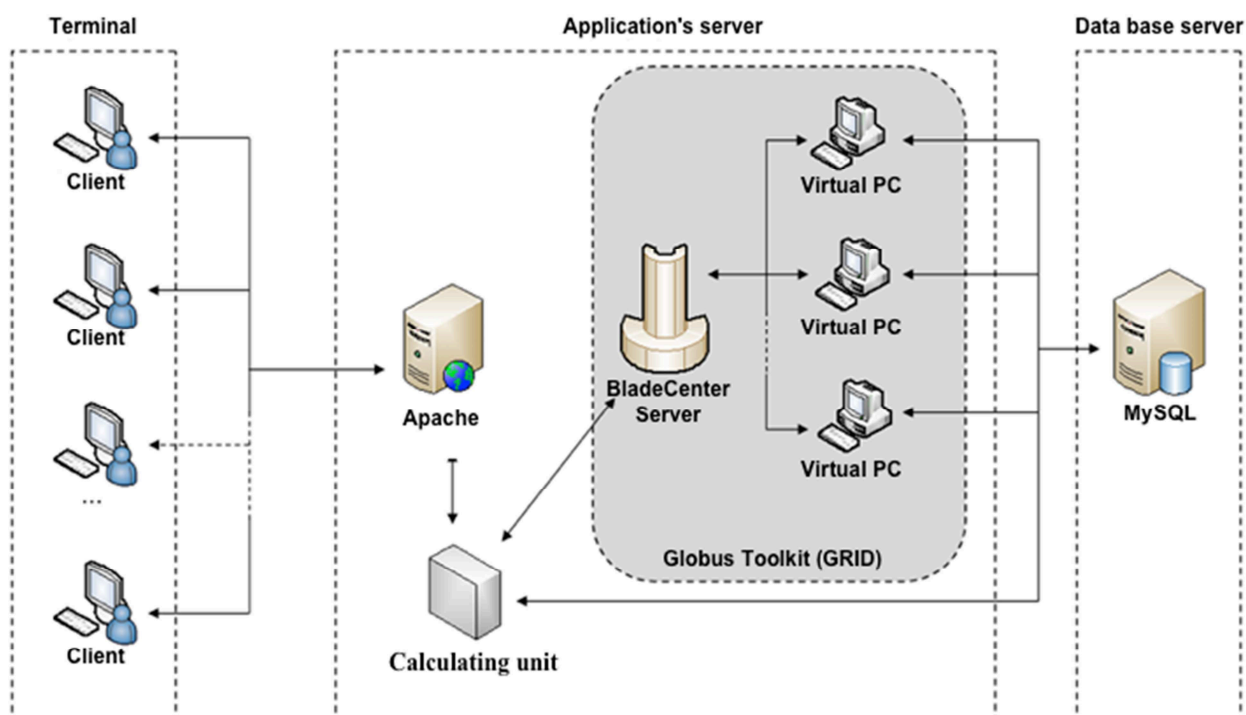


Fig. 1. Architecture of the software application

The application server is designed for performing all calculations required from the system, interconnecting the interface component and the database server as well as for performing distributed computing.

To perform calculations, a calculating unit is allocated. It is designed for the following:

- modelling of phase transformations at given isothermal tempering;
- 3D imaging;
- calculating the kinetic parameters of the material on the basis of entered diagrams of austenite decomposition under isothermal conditions;
- computing data of thermokinetic diagrams;
- digitizing isothermal and thermokinetic diagrams;
- predicting material structure and calculating stereometric parameters.

The database server is designed for the application user to interact with the information stored on a database. It helps to save all information entered by the user and the results of computations performed. This server enables users to view the results of previous computations.

For client Windows application and Web server to work properly the Internet Direct (Indy) library is used. It allows implementing specific server logic which is absent in standard Web servers. [3, 4]. The Indy library is comprised of a component kit divided into three groups: client (Indy Client), server (Indy Servers), and auxiliary (Indy Misc) components. Indy client components are developed by using socket technology. The client-side socket requests server connection. If the connection is established, the client and the server can exchange messages. The exchange takes place via HTTP. The Indy library components are used to support one of the main network protocols - TCP (Transmission Control Protocol).

Constituents of the Calculating Unit

The calculating unit comprises a set of different modules performing specified functional load of the software application. The description of specific features of each module is provided below.

The module of modelling of phase transformations at given isothermal tempering. Within the framework of this module the computer modelling of structural transformations in steel is performed by the entered initial data: the size of the area under study, the parameters of a simulation model, the tempering temperature.

The module output is a tabulated isothermal curve in the "time – austenite volume fraction" coordinates as well as data on the size, type, location of the structural constituents throughout the transformation process.

The module of 3D imaging. The module is designed to demonstrate 3D animated modelling. The module outputs are shown in Figures 2, 3. Figure 2 shows the model condition at the intermediate time point, it is seen how the structural constituents of different phases arise and grow. Figure 3 shows the final state of the model that reflects the end of imaging.

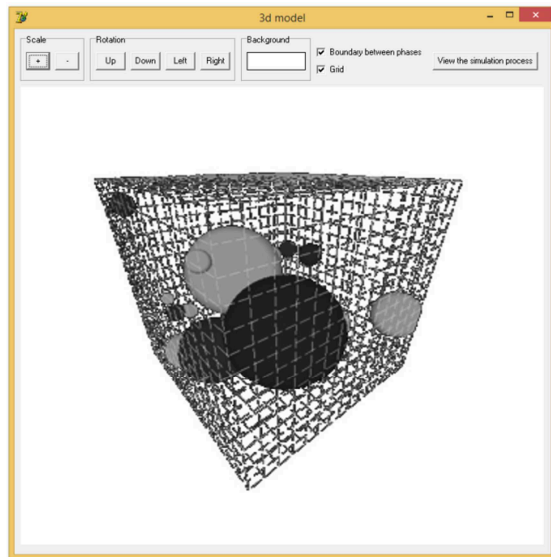


Fig. 2. Model condition at the intermediate time point

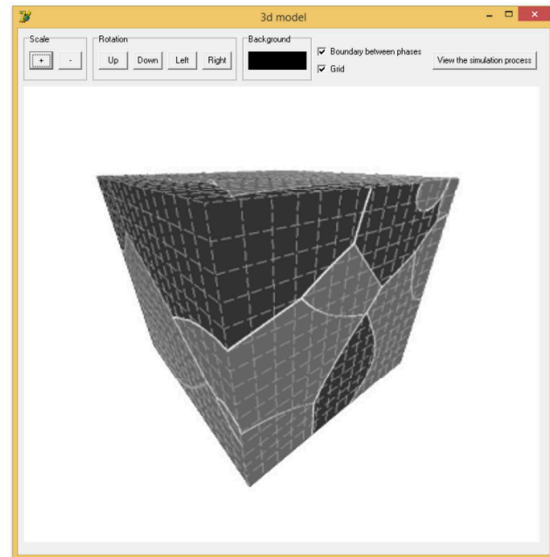


Fig. 3. Final state of the model

The module enables the 3D model to rotate arbitrarily. It allows changing the scale and colour as well as showing or hiding the gridlines and displaying the boundaries between the structural areas. 3D imaging is constructed using the Open Graphics Library (OpenGL) [5].

The module of calculating the kinetic parameters is designed to calculate model kinetic parameter values. Two options for data entry are provided: by digitizing diagrams or setting lists of values (time - fraction of decomposed austenite) for each temperature tempering.

After entering all required data the calculation algorithm starts using the modified Nelder-Mead method [6].

The module of digitizing isothermal and thermokinetic diagrams. The module provides access to the following functions:

- Data handling.
- Specifying the scale of data presentation (linear or logarithmic).
- Analyzing the image spectrum to separate the objects on the diagram from the background.
- Digitizing graphs.
- Exporting the received points to the XML file.

The system allows saving the initial diagrams and the results of digitizing as well as editing the digitized graphs. Figure 4 shows the form of digitization module.

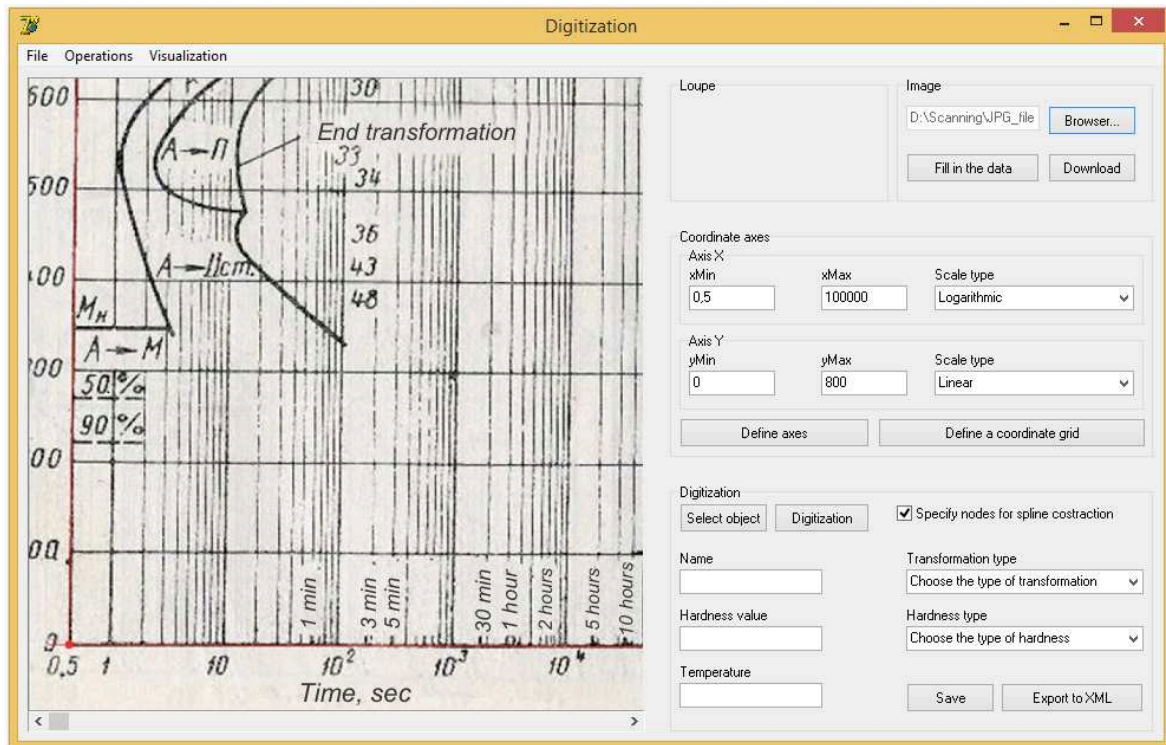


Fig. 4. Form of digitization application

Figure 5 shows two images: the initial and digitized isothermal diagrams.

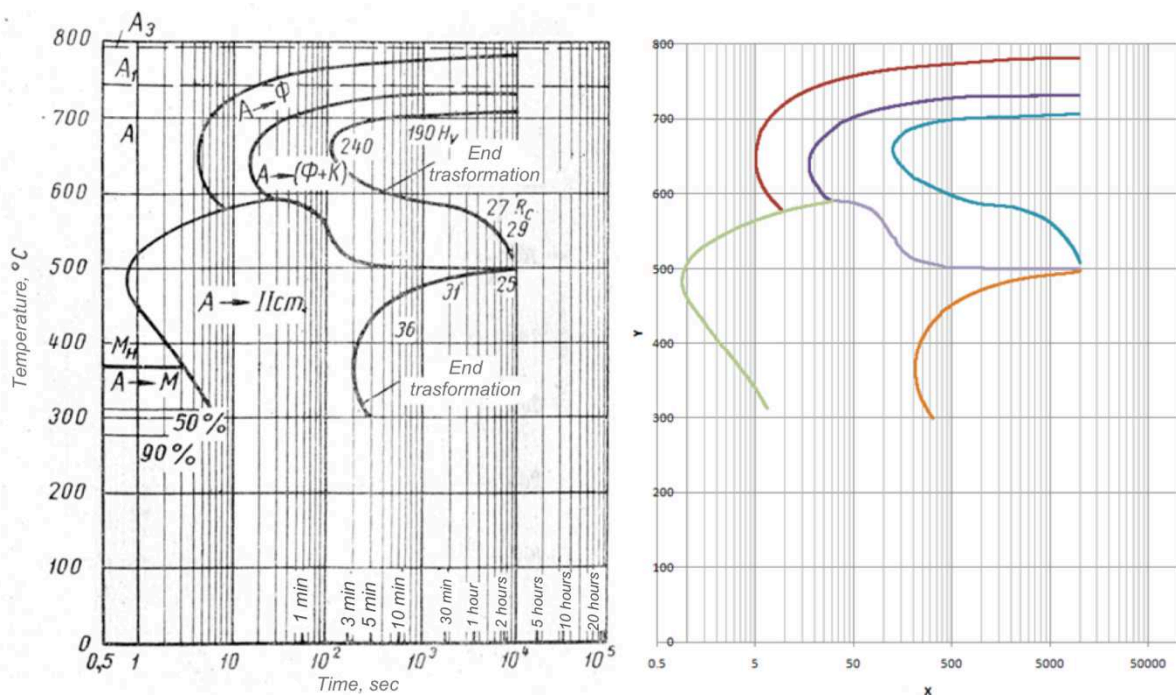


Fig. 5. Comparison of digitization results

The module of predicting steel structure. This module has the following functional capabilities:

- plotting a metallographic structure in the plane (microsection) based on the results of phase transformation modelling under isothermal tempering;
- 3D imaging of the plotted microsection;
- estimating stereometric parameters of the process modelled.
- Server side

The server side employs Java JSP technologies and auxiliary classes – servlets. The classes of servlet are located on the web server and they control the entire system (Figure 6). Their task is to provide the access to the system functions based on user requests via web interface.

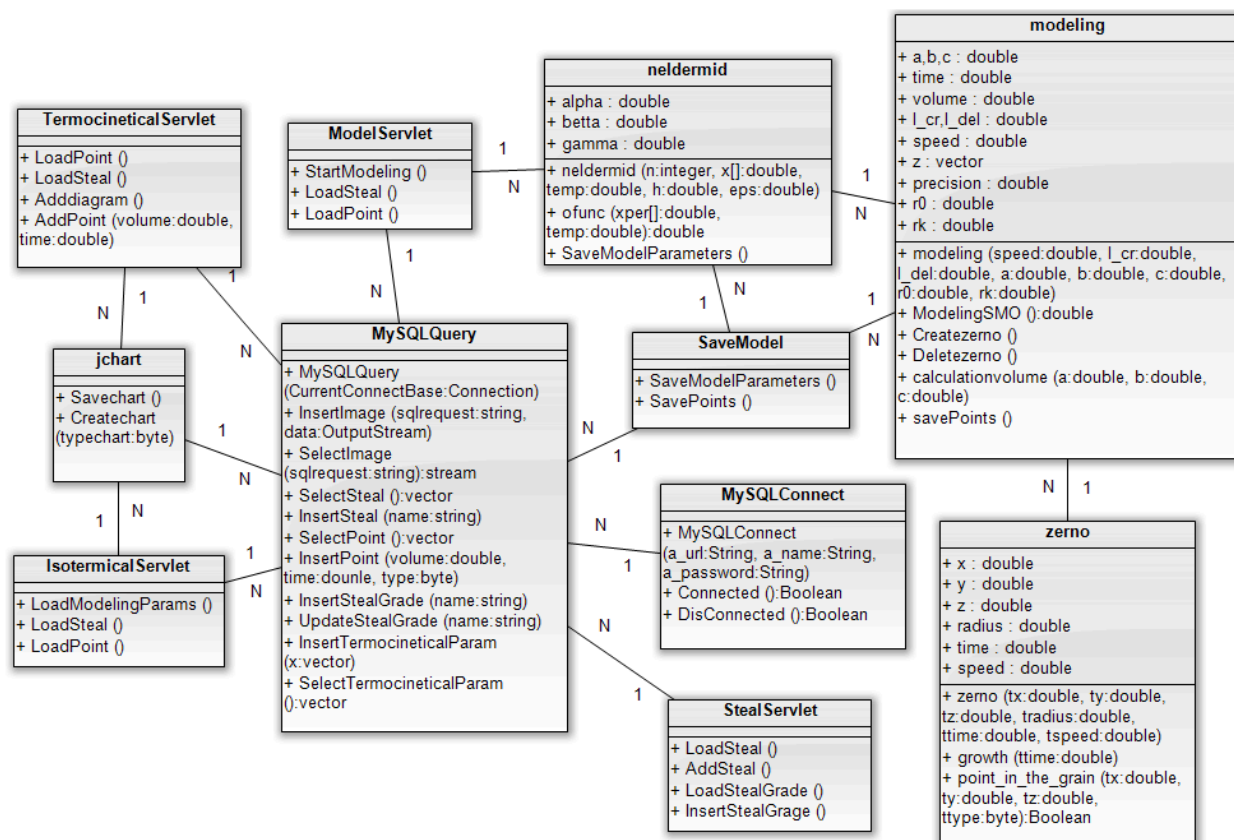


Fig. 6. Class diagram

The StealServlet class is designed to view and edit steel grades. The LoadSteal method allows downloading steel information from the database. The AddSteal method is designed to add information on a new grade of steel. The LoadStealGrades and InsertStealGrade methods allow downloading and adding information on the chemical composition of steel. The TermocineticalServlet class allows adding the initial data of thermokinetic experiments by the AddPoint method. The IsotermalServlet class has similar capabilities concerning isothermal diagrams. The ModelServlet class starts modelling using the StratModeling method.

The classes of Web pages interact with the database by two classes. The MySQLQuery class is designed for the interaction between the system and database. This class is used to save the initial data and experimental results. The connection with the database is established using the MySQLConnect class that allows establishing and breaking connection with the BDMS.

The Modeling class instances are responsible for modelling. The modified Nelder-Mead method is implemented by the Nelder-Mid class. After completing all experiments the SaveModel class collects the results of modelling and adds them to the database.

The Description of Using the Distributed Computation Technology

Since the computation of kinematic parameter values involves computer-intensive, hardware and time resources, it is reasonable to apply the distributed computation technology [7, 8].

For the purpose to provide distributed computing, the Grid system is used [9, 10]. It is a cluster of the computers connected through the network, one of which serves as a host and others are secondary. To implement such a GRID system, Globus Toolkit 4 is used [11, 12]. It is a set of software components implementing the required parts of the GRID architecture

In addition to the Globus Toolkit, to provide a convenient access to the computing resources and replace a local task scheduler, TORQUE is used as a manager of the computing resources. The GRID system is implemented through the IBM Blade-Center server that allows virtualizing the computers classified as the secondary ones [13].

Conclusions

1. The recognition algorithm of raster images on the isothermal and thermokinetic diagrams has been proposed. It allows using a semi-automatic mode of initial data entry.
2. The imaging algorithm of the material structure after heat treatment and the algorithm of calculating its parameters have been developed.
3. The architecture of the software application has been developed where through the GRID system the chart of distributed computation of simulation model kinematic parameters has been implemented. It allows considerable reducing time expenditures during computer experiments.
4. The developed software application allows the user to model phase transformations under isothermal conditions for a specified steel grade at arbitrary tempering temperature.

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Structural-Phase State and Strength Properties of Pressure-Synthesized Ni_3Al Intermetallic Compound

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Keywords: intermetallic compound, self-propagating high-temperature synthesis, the structural phase state, grain size, strength of the intermetallic compound

Abstract. The article studies dependences of grain size in Ni_3Al intermetallic compound synthesized under pressure in $3\text{Ni}+\text{Al}$ powder mixture in conditions of bulk exothermal reaction upon pre-pressure acting on the powder mixture and upon a delay time of applying pressure to a high-temperature synthesis product. It is proved that an increase in the pre-pressure on the parent powder mixture reduces the grain size, and an increase in the delay time increases the grain size in the synthesized intermetallic compound. Reducing the grain size from 10 to $1.75\mu\text{m}$ increases the strength of the intermetallic compound under pressure from 336 to 482 MPa (1.4 times).

Introduction

Self-propagating high-temperature synthesis (SHS) of intermetallic compounds initiated by thermal explosion of powder mixtures of parent elements is an alternative technological development in producing advanced intermetallic alloys and composite materials for structural and instrumental purposes [1-5]. Thermal and physical conditions of the intermetallic compound bulk intermetallic synthesis reaction in a parent powder system ensure simultaneous phase transformations in the entire volume of the heat-sensitive powder mixture and enable consolidation of single structural fragments of the reaction products at the point the product reaches the homogeneous structural-phase state caused by external pressure.

Due to a high speed of phase changes under conditions of bulk exothermal reaction of intermetallic synthesis in the powder mixture of parent elements it is extremely difficult to control the generation of structural phase state and strength properties of a pressure-synthesized intermetallide. This is especially important when developing technologies of high-temperature synthesis of articles from intermetallic compounds and intermetallic alloys by means of power compaction of high-temperature synthesis products. The main drawback of the technologies suggested in [6,7] which does not allow controlling the process of structural-phase state generation of the synthesized intermetallic compounds consists in the fact that the intermetallic high temperature synthesis is initiated and realized under the condition of powder blank front combustion which prevents from achieving a homogeneous structural-phase state in the whole of the synthesized product by the moment the pressure is applied and as a result the predetermined grain size can't be obtained in the whole volume of an intermetallide synthesized under pressure. A homogeneous structural-phase state can be obtained only when we synchronize the processes of the bulk exothermal reaction of intermetallic synthesis in the parent powder mixture and the compaction of the synthesis products.

This article presents the research results of the influence of high temperature synthesis under pressure in conditions of bulk heating of powder mixture $3\text{Ni}+\text{Al}$ on grain structure size and strength characteristics of Ni_3Al intermetallide synthesized under pressure.

Research Methods and Materials

A high-temperature synthesis of Ni_3Al intermetallide under pressure was carried out with the help of a test stand based on an automated hydraulic press equipped with high-frequency generator for heating a steel press-mold until a $3\text{Ni}+\text{Al}$ powder mixture compact self-ignites (Nickel grain size being $1\ldots 3\text{ }\mu\text{m}$, and that of aluminum being $5\ldots 10\text{ }\mu\text{m}$). The test stand was equipped with a high frequency generator for continuous heating of the steel press-mold, a temperature recorder of powder compact heating, a digital monometer for recoding pressure in the hydro-system of the main press cylinder and a timer of the work pressure in the main press cylinder (Fig. 1).

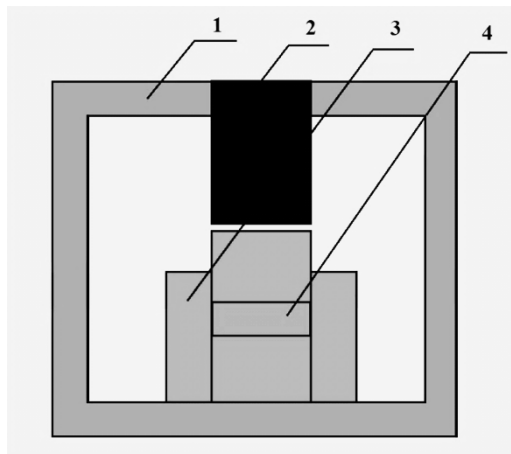


Fig. 1. Block diagram of a test stand for high-temperature synthesis of an intermetallic compound under pressure during continuous heating of a powder mixture of the parent elements in a steel press-mold: 1- working space of the hydraulic press, 2 - main press cylinder, 3 - steel press-mold, 4 - powder compact.

The structural phase state of the synthesized intermetallic samples was investigated by optical metallography (Neophot 32), X-ray diffraction analysis was carried out on a DRON-7 X-ray diffractometer in $\text{CoK}\alpha$ radiation at an accelerating voltage of 35kV and a current of 20mA. The tensile strength of the intermetallic samples was examined by a LFM-125 machine (walter + bai ag Testing Machines, Switzerland) at a strain rate 0.2 mm/s (for dimensions of the working part of the samples $10 \times 3 \times 1.2\text{ mm}$).

Results and Discussion

Fig. 2 shows a "pressure-time" diagram for high-temperature synthesis of an intermetallic compound under pressure in conditions of continuous heating of the parent powder mixture. High-temperature synthesis of Ni_3Al intermetallic compound under pressure goes through the following sequence of synthesis stages:

- (1) a preload is applied to the $3\text{Ni} + \text{Al}$ powder in the press form;
- (2) heating the mold with high-frequency currents causes self-ignition of the powder mixture, the value of the preload on the powder mixture is reduced to a certain minimum;
- (3) a pre-programmed timer activates the press working pressure in accordance with the preset time of the product compaction delay.

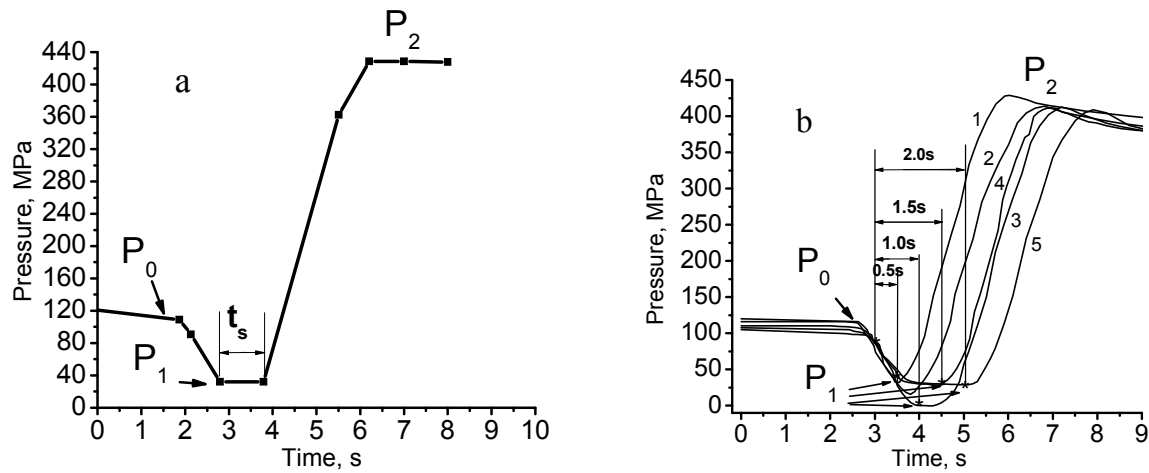


Fig. 2. Pressure-time diagram of the high-temperature synthesis of Ni_3Al intermetallic compound under pressure in continuous heating of the $3\text{Ni} + \text{Al}$ powder mixture (a) and experimentally obtained high-temperature synthesis diagrams of the Ni_3Al at various pre-pressures acting on the initial $3\text{Ni} + \text{Al}$ powder mixture and delays of the product compaction time t_s (b):

P_0 is the value of the pre-pressure, P_1 is the value of the minimum pressure on the powder mixture after it is heated to self-ignition, P_2 is the value of the compacting pressure on the high-temperature synthesis product.

X-ray phase analysis of Ni_3Al samples at different values of the preliminary pressure on the parent powder mixture and delay in application time of the high-temperature synthesis product compacting pressure after self-ignition of the powder mixture showed that in all process variants we obtain Ni_3Al monophase nickel aluminide as the product of high-temperature synthesis (Fig.3, 4).

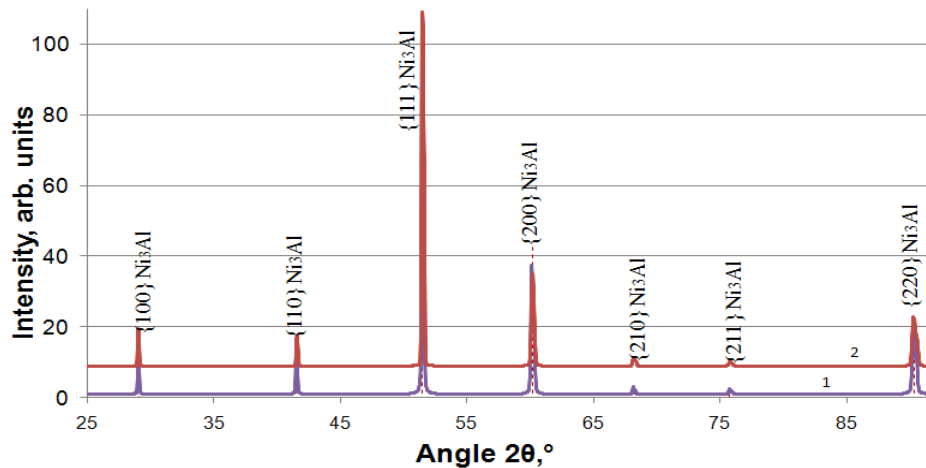


Fig. 3. Diffractograms of Ni_3Al samples synthesized under pressure and continuous heating in a steel mold at initial pressures on a nickel and aluminum powder compact ($3\text{Ni} + \text{Al}$) of 33 MPa (diffractogram 1) and of 115 MPa (diffractogram 2) without delay in time before compaction of the synthesis product.

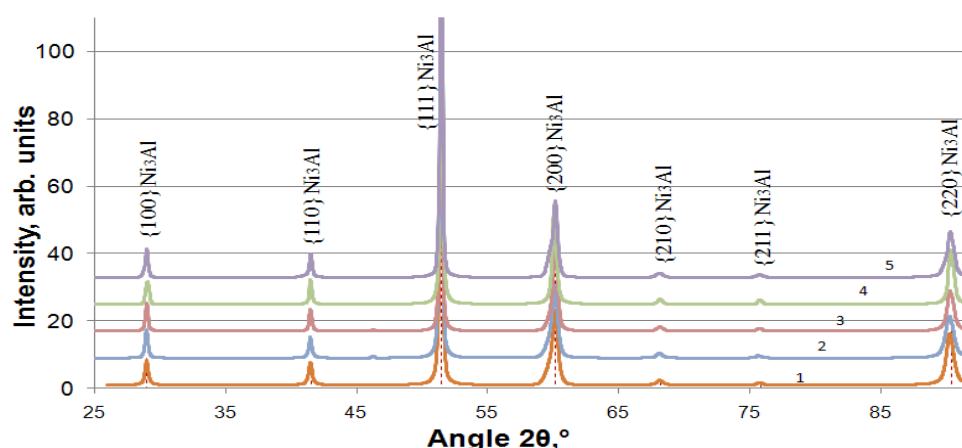


Fig. 4. Diffractograms of Ni_3Al samples synthesized under pressure at different delays in time of application of pressure to the high-temperature synthesis product: 1 - 0 s, 2 - 0.5 s, 3 - 1.0 s, 4 - 1.5 s, 5 - 2.0 s.

The change in the initial pressure affects the dimensionality of the grain structure of Ni_3Al intermetallide synthesized under pressure. Fig. 5 is a general view of the grain structure of the pressure-synthesized Ni_3Al intermetallide at various pre-pressures acting on the initial powder compact in a mold without delay in time till the compaction of the synthesis product. As the pre-pressure on the powder compact is increased, the grain size in the synthesized intermetallic compound decreases. Fig. 6 shows the quantitative dependencies of grain size and hardness of the pressure-synthesized Ni_3Al intermetallide upon the pre-pressure acting on the parent powder mixture.

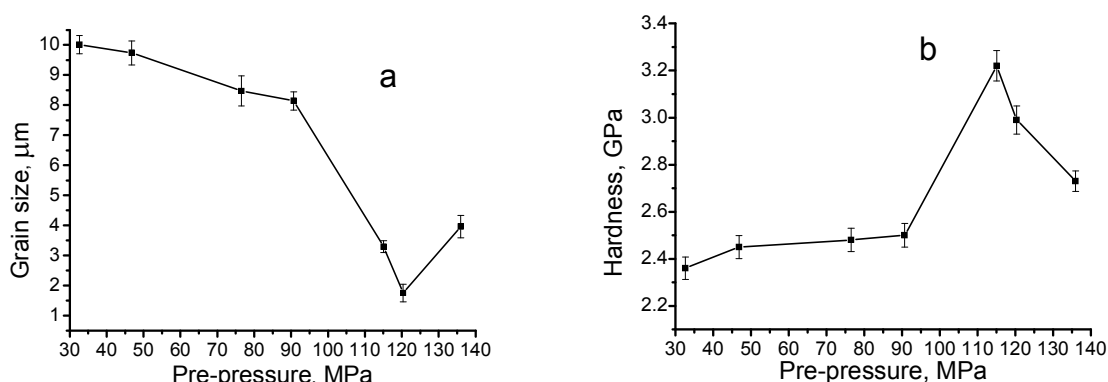


Fig. 5. Dependences of the grain size (a) and hardness (b) of the pressure-synthesized Ni_3Al intermetallide upon the pre-pressure acting on the parent powder mixture at constant parameters of high-temperature synthesis under pressure.

As the pre-pressure on the powder mixture in the mold is increased from 33 to 120 MPa before its heating to self-ignition, the grain size in the intermetallic compound synthesized under pressure decreases from 10 to 1.75 μm , while the hardness of the intermetallic compound increases from 2.4 to 3.2 GPa. With a further increase in the pre-pressure on the powder mixture to 140 MPa, the grain size increases up to 4.0 μm and correspondingly the hardness of the intermetallic compound decreases to 2.7 GPa.

The opposite changes in the grain size and the hardness of the intermetallic compound synthesized under pressure take place when the delay time is increased till compaction of the synthesis product takes place. As the delay time increases, the grain size grows and the hardness of the intermetallic compound decreases (Fig. 6)

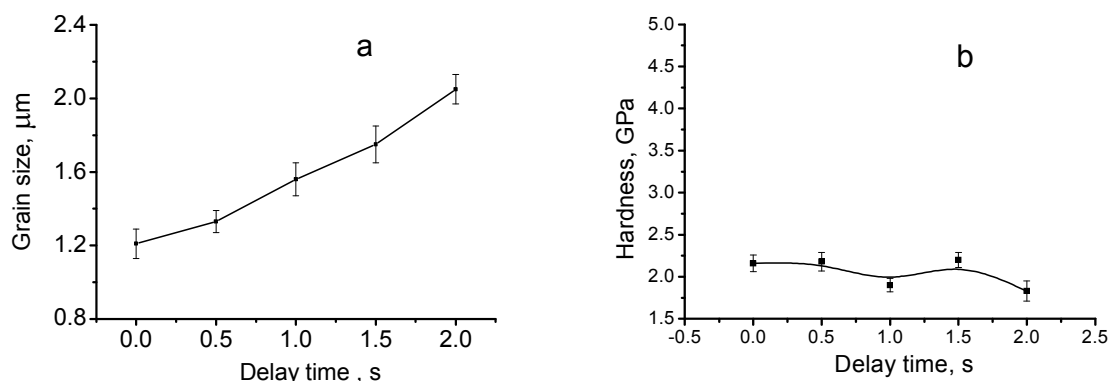


Fig. 6. Dependence of the grain size (a) and hardness (b) of the pressure-synthesized Ni_3Al intermetallide upon the delay time until compaction pressure is applied to the synthesis product.

Reduction of the grain size in the pressure-synthesized intermetallide when increasing the pre-pressure on the parent powder mixture is due to an increased number of contacts of dissimilar particles in the powder mixture. These particles determine the number of nuclei generating the intermetallic compound and in the process of crystallization of the high-temperature synthesis product, the number of grains in the emerging polycrystalline structure. The increase in the exposure time of the synthesis product at the melt crystallization temperature till the moment of its compaction determines the conditions for a more complete mutual diffusion processes between individual embryos of the grain structure which ensure the coalescence of individual embryos and the generation of larger grains.

Changes in the grain size of the pressure-synthesized intermetallide Ni_3Al - significantly influence its strength. Fig. 7 shows the dependence of the tensile strength of an intermetallic compound at room temperature – increasing the grain size from 1.75 to 10 μm reduces the strength of the intermetallic compound by 1.4 times.

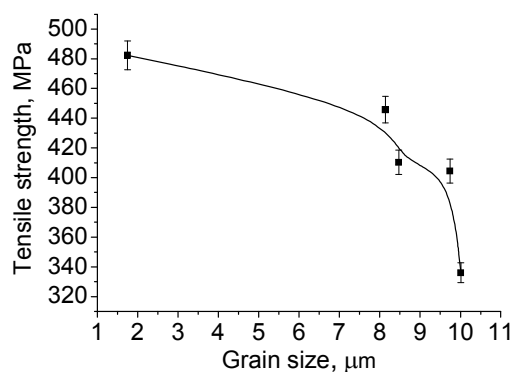


Fig. 7. Dependence of tensile strength at room temperature of the pressure-synthesized Ni_3Al intermetallide upon the grain size of its polycrystalline structure.

Conclusion

The pre-pressure on the parent powder mixture $3\text{Ni} + \text{Al}$ and the dwelling time of the high-temperature synthesis product till its compaction are the key parameters of Ni_3Al intermetallic compound high-temperature synthesis under pressure. These parameters determine the grain size in the synthesized intermetallic compound polycrystalline structure.

While an increase in the pre-pressure acting on the powder mixture results in a decrease in the grain size of the pressure-synthesized intermetallic compound, an increase in the dwelling time of the synthesis product till its compaction increases the grain size. Changes in grain size have a significant effect on the strength of the synthesized intermetallic compound. When the grain size decreases from 10 to 1.75 μm tensile strength of the intermetallic compound at room temperature increases by 1.4 times.

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Structure and Properties of the Surface Layer of a Wear-Resistant Coating on Martensitic Steel after Electron-Beam Processing

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Keywords: steel, layer, structure, properties.

Abstract. The paper reports electro-contact welding on Hardox 450 steel with C-V-Cr-Nb-W flux-cored wire. Supplementary irradiation by intense pulsed electron beam was carried out to improve mechanical properties. Micro- and nano-hardness, Young modulus and tribological parameters of the modified surface were tested mechanically. It is pointed at the significant increase in the friction coefficient because the surface layer fractures and particles of the surfaced layer are involved in the process of friction. Using the methods of optical and scanning microscopy a great number of micropores were detected both on the irradiated surface and through the surfaced layer modified by intense pulsed electron beam. It is demonstrated that electron-beam processing of the deposited layer surface is the reason for occurrence of multi-layer structure. According to measurements it was determined that the modified (surface and transition) layers are 0.3 to 0.5 μm on overage. It was also found out that irradiation of the surfaced metal leads to significant refining of structural elements because of ultrahigh speeds of crystallization and further cooling down of the modified layer. The phase composition of the surfaced metal modified by pulsed electron beam is explored. Niobium carbide (NbC) is reported to form in the surface layer.

Introduction

Parts of machines and mechanisms are operated at full capacity in different industries (metallurgical plants, mining, oil-producing, road-repairing and other enterprises). Laser, plasma, electron-beam, electro-arc etc. methods of surfacing are used to protect metals and alloys from various types of environmental loads – wear, corrosion, high temperatures, static and dynamic stresses [1-12]. Coatings with a number of characteristics, e.g. wear and corrosion resistance are of great practical importance. Such valuable characteristics can be provided via coating the surface of a product with layers of metal with advanced operational properties. When welding an alloy is formed, so the surfaced metal is a single whole with the substrate metal. Welding is not a cost-intensive but a high efficient method to repair and protect products. At the same time technological defects of welding can be eliminated quite well with the help of irradiation by intense pulsed electron beam. One of the promising method of modification of near surface layer structure of products from different materials is electron-beam pulse processing. In comparison with a widely spread laser processing the electron-beam technology has wide possibilities of control and regulation of amount of supplied energy and differs in locality of energy distribution in near surface layer of the materials being processed and a high efficiency. Ultrahigh speeds (108-1010 K/s) of heat to temperature of melting followed by cooling of thin surface layer of the materials (10^{-7} - 10^{-6} m), very little (10^{-6} - 10^{-3} s) time of high temperature effect, formation of limiting gradients of temperature (up to 107-108 K/m), ensuring the cooling of near surface layer at the expense of heat removal into the basic volume of the material at the rate of 104-109 K/s, create conditions for the formation of amorphous, nano- and submicrocrystalline structure in near-surface layer. These appreciable changes of structure phase

state of surface layers result in the significant improvement of physico-chemical and strength properties of the materials unattainable at traditional methods of surface processing [13-19]. Hence, the paper aims at exploring the structure, mechanical and tribological properties of the layer formed on Hardox 450 steel via electro-contact welding with iron-based C-V-Cr-Nb-W wire and additionally irradiated by intense pulsed electron beam.

Material and Methods of Research

As in [20-22] we used Hardox 450 steel ((w.%): 0.19-0.26% C; 0.70% Si; 1.6% Mn; 0.025% P; 0.010% S; 0.25% Cr; 0.25% Ni; 0.25% Mo, 0.004% B; balance – Fe) as the substrate material. The flux-cored wire PP-2 (w. %): 1.4% C; 1.0% V; 7.0% Cr; 8.0% Nb; 1.2%W; balance – Fe was used in the process of welding. MIG/MAG welding was carried out in the shielding gas containing Ar 82 %, CO₂ 18%, welding current 250 – 300 A, voltage 30 – 35 V.

The deposited layer was modified by means of irradiation with high-intense electron beam, the facility SOLO [13, 14, 19] was used in the mode of melting and high-speed crystallization. Irradiation was carried out in the following conditions: energy density of electron beam in the pulse – 30 J/cm²; pulse duration – 200 μs; number of pulses – 20 at the first stage; energy density of electron beam in the pulse – 30 J/cm²; pulse duration – 50 μs; number of pulses – 1 at the second stage. The conditions of irradiation were determined with the help of calculating the temperature field, which is formed in the surface layer of the material when irradiating in the single-pulse mode. To carry out mechanical tests of the modified surface micro-hardness (micro-hardness measuring device HVS-1000, Vickers method) was measured; the indenter was loaded 0.2 N and 0.5 N, as well as nano-hardness and Young modulus were determined (ultra and micro-hardness measuring device Shimadzu DUH-211S; the indenter was loaded 50 mN). Tribological tests of the modified surface were conducted using the tribo-test device “CSEM Tribometer High Temperature S/N 07-142”, CSEM Instruments; the speed of wear was assessed according to the cross-sectional area of a wear track with the 3D-profilometer MICRO MEASURE 3D station by STIL. The structure of the material was explored via analyzing thin sections cut off perpendicularly to the surface to be modified. The defect structure of the material was explored by SEM methods (scanning electron microscope “SEM-515 Philips”) and optical microscopy (microscope “Metallographic micro-visor” μVizo - MET-221”). The elemental composition of the surface layer was explored with the help of X-ray microanalysis. X-ray diffraction was used to analyze a phase composition of the surface layer (diffractometer XRD-7000s, Shimadzu).

Results and their Discussion

As shown in the study on strength properties of the irradiated surface, micro-hardness of the surface layer is decreased by ≈1.8 times after modification, as compared with that of the not modified layer and is almost similar to micro-hardness of the initial steel. At the same time it is pointed at the approximate increase in Young modulus by 1.1 times.

Tribological tests of the deposited layer modified by pulsed electron beam demonstrate that its wear resistance was significantly improved (by ≈32.5 times) in conditions of friction with a ball made of hard alloy VK6. The friction coefficient was as ≈1.2 times high as that of Hardox 450 steel in the initial state.

As identified in wear resistance tests, there are two stages in the behavior of the friction coefficient (Fig. 1). At the first quite short stage of the tests the friction coefficient is as ≈1.3 times low as that of the initial steel (Fig. 1, segment I); at the second stage it exceeds the friction coefficient of the initial steel by ≈1.3 times on average (Fig. 1, segment II). We suppose that the significant increase in the friction coefficient at the second stage of the tests is possible as the surface of the deposited layer gets destructed and its particles are involved into the process of friction.

The phase composition and defect substructure of Hardox 450 steel and those ones of the deposited layer irradiated by the intense pulsed electron beam were explored using the methods of scanning electron diffraction microscopy of thin foils. The findings of these studies point out that Hardox 450

steel is a multiphase material, the main phases of which are α -phase and iron carbide. α -phase is represented by two morphologically different types of structure, i.e. packet martensite and sub-grains. The basic type of the steel structure is packet martensite; the crosscut sizes of its crystals vary 150 to 200 nm. In the volume and on the boundaries of martensite crystals, as well as sub-grains of α -phase there are particles of the second phase. When indexing the micro-electron diffraction patterns taken on these zones of the material it was revealed that these particles are iron carbide (Fe_3C). Taking into consideration the structural morphology of α -phase and phase composition, it can be concluded that Hardox 450 steel had been hardened before the deposited layer was formed (packet martensite was generated) and tempered (particles of iron carbide and zones with sub-gran structure occurred).

The surface of the deposited layer modified by intense pulsed electron beam is destructed when wear resistance testing due to a great number of micro-pores. Fig. 2 illustrates typical structures of the welded surface irradiated by intense pulsed electron beam. Numerous micro-pores are distinctly seen, their sizes vary in the micrometer range (Fig. 2, a). Electron-beam processing of the welded surface when melting the surface layer results in formation of a poly-crystalline structure, the grain sizes of which vary $0.2\text{ }\mu\text{m}$ to $2\text{ }\mu\text{m}$ (Fig. 2, b).

A typical structure of the etched thin section is given in Fig. 3. First of all, it is worth to mention that there are pores both on the irradiated surface (Fig. 2, a) and through the deposited layer (Fig. 3, a, pores are indicated by white and black arrows). A multi-layered structure is formed when irradiating the deposited metal by the intense pulsed electron beam (Fig. 3, b). There is a surface layer (Fig. 3, b, layer 1), an intermediate layer (Fig. 3, b, layer 2) and a layer of thermal impact (Fig. 3, b, layer 3).

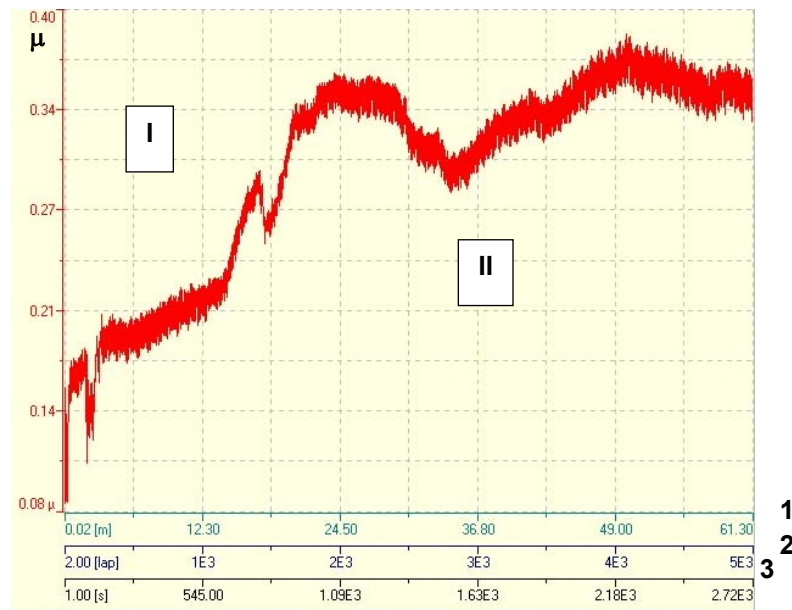


Fig. 1. The curve of friction coefficient vs. friction track length (1), number of revolutions of the sample when friction testing (2), test duration (3).

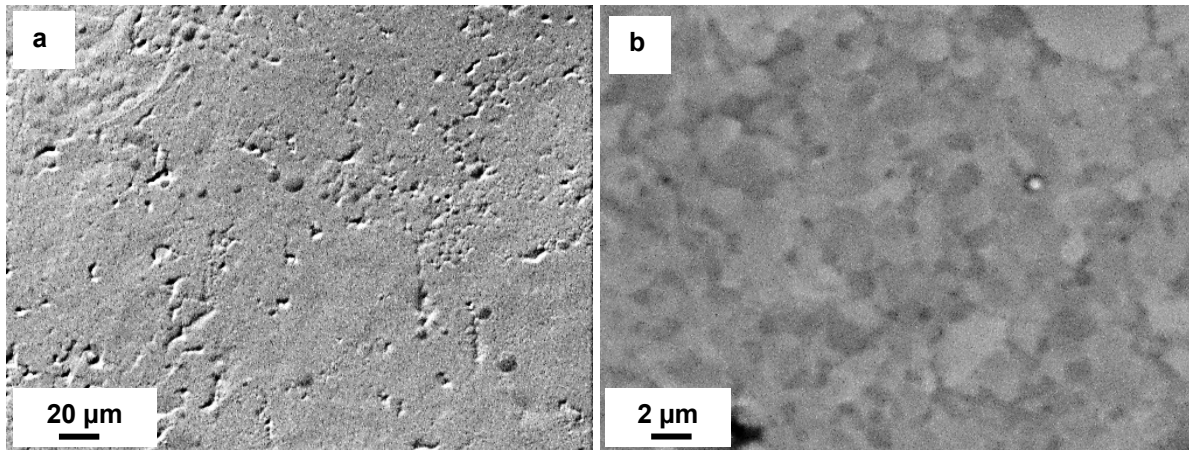


Fig. 2. Electron microscopic image of the surface layer formed on Hardox 450 steel via electro-contact welding with PP-2 wire and irradiated by electron beam. SEM.

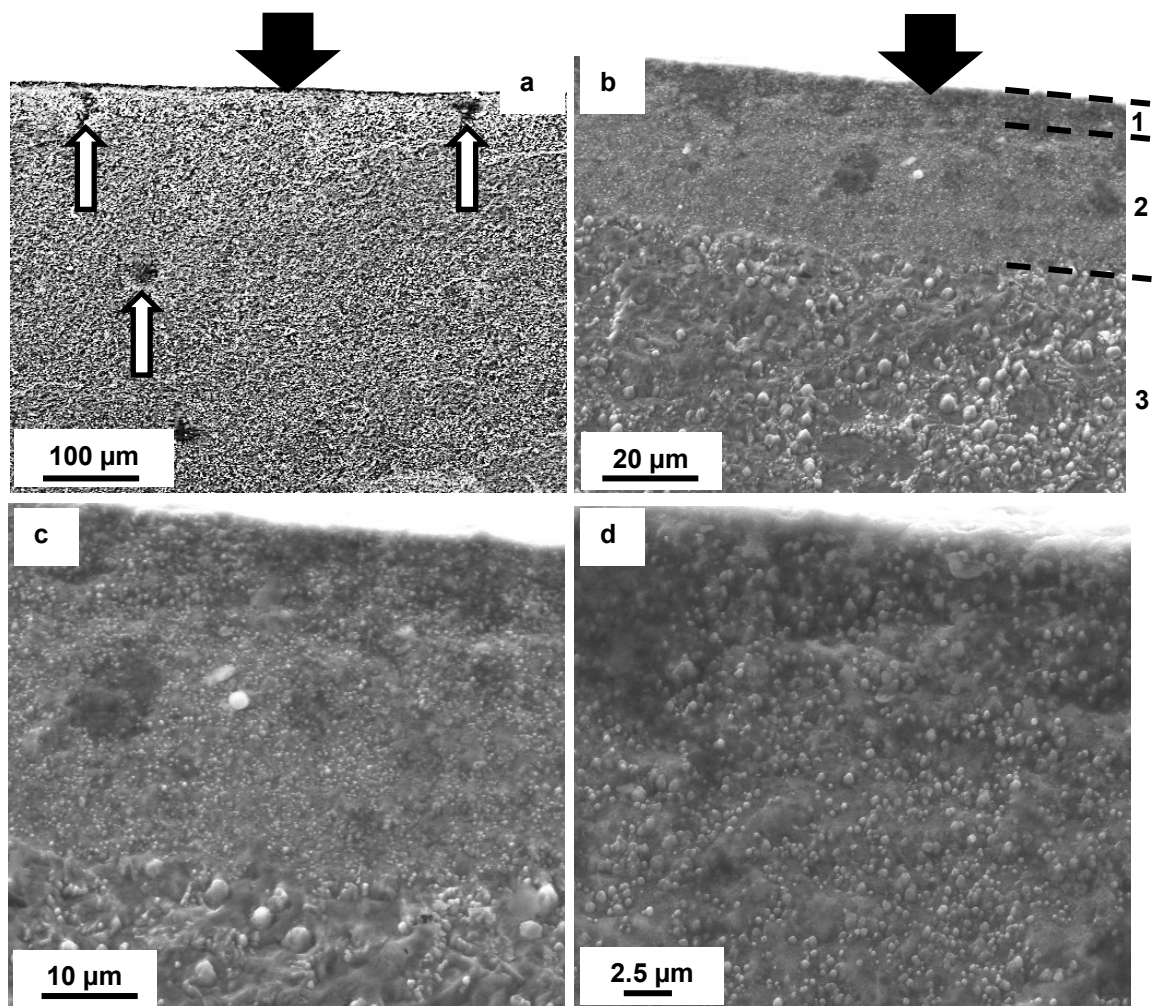


Fig. 3. The cross section of the deposited layer formed on Hardox 450 steel via electro-contact welding by iron-based C-V-Cr-Nb-W wire and exposed to electron beam irradiation. SEM. Black arrows in (a) and (b) point at the irradiation surface; black and white arrows in (a) point at the micro-pores; (b) the figures designate the layers formed in the surface layer of the deposited metal in the process of electron beam irradiation.

As seen, the structural elements of the surface and intermediate layers are far smaller than inclusions of the carbide phase in the underlying metal structure (Fig. 3, b-c, layer 3). It was estimated that the average sizes of the structural elements of the modified (surface and intermediate) layers vary 0.3 μm to 0.5 μm (Fig. 3, d). The average sizes of the structural elements throughout the main volume of the deposited metal are up to 5 μm (Fig. 3, b). Ultrahigh speeds of crystallization

and further cooling of the modified layer evolving in conditions of pulsed electron-beam processing are the reasons of the identified transformations [23].

A typical X-ray diffraction pattern is given in Fig. 4, which was made on the surface layer of the deposited metal irradiated by intense electron beam. It is found out that niobium carbide (NbC) occurs in the surface layer, its relative content is 0.3 wt. %. As seen in the elemental composition of the deposited material and corresponding phase equilibrium charts [23, 24], diffraction peaks indicated by the dashed arrows are in the nonstoichiometric carbide phase.

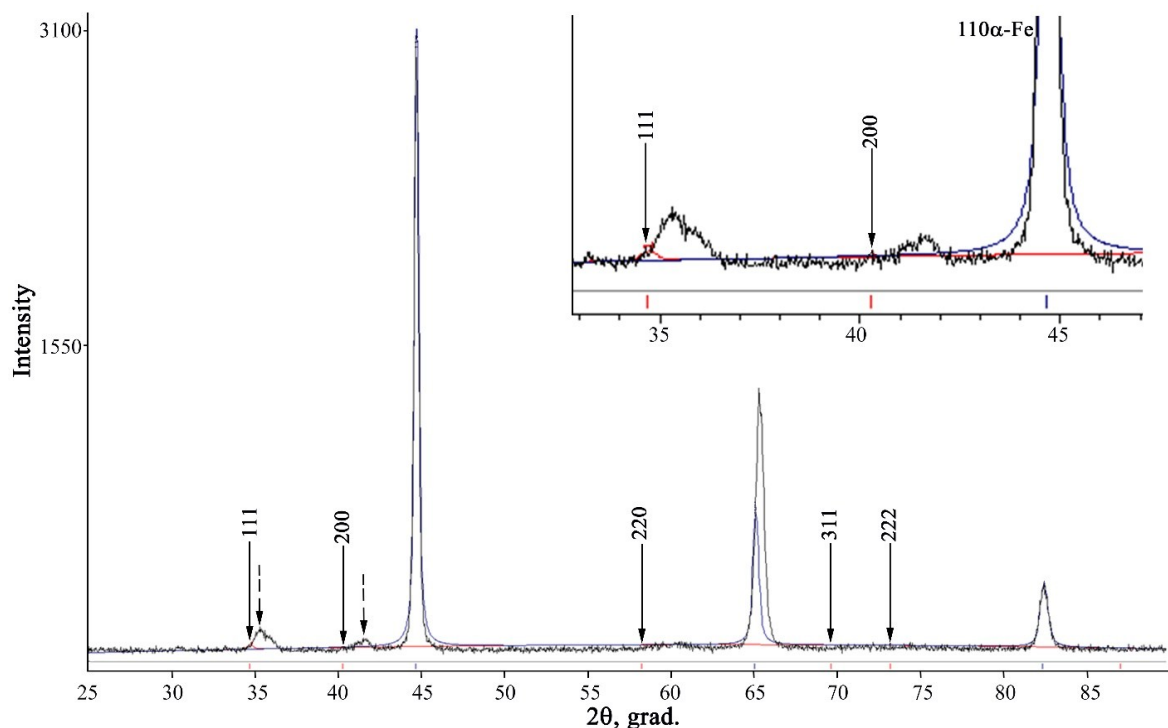


Fig. 4. A segment of the X-ray diffraction pattern obtained on the surface of the deposited layer irradiated by pulsed electron beam; indicated diffraction peaks of niobium carbide (solid arrows) and nonstoichiometric composition of the carbide phase (dashed arrows).

Conclusions

It is indicated that electron-beam processing of the hardened layer on Hardox 450 steel deposited via electro-contact welding with iron-based C-V-Cr-Nb-W wire results in multifold improvement of wear resistance, Young modulus and friction coefficient, as well as causes the decrease in micro-hardness of the surface modified layer as compared with the substrate (Hardox 450 steel). It is found out that advanced tribological properties of the deposited layer modified by electron beam are possible due to the multilayered structure; its hardening is the result of tempering. The tempering effect is associated with refining of structural elements when electron-beam processing and inclusions of the carbide phase (niobium carbide and nonstoichiometric carbide phase).

Acknowledgements

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Studying the Influence of Tungsten and Chromium Additives in Flux Cored Wire System Fe-C-Si-Mn-Mo-Ni-V-Co on Surfaced Metal Properties

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Keywords: flux cored wire, surfacing, microstructure, hardness, wear resistance, grain size, nonmetallic inclusions impurity.

Abstract. In the laboratory conditions were manufactured flux cored wire system Fe-C-Si-Mn-Mo-Ni-V-Co samples, studied the influence of tungsten and chromium appending in surfacing wire charge. Conducted metallographic researches of surfaced metal: microstructure evaluate, grain size, nonmetallic oxides inclusions impurity. Defined hardness and wearout speed of the deposited layers. Conducted evaluation of the chemical composition influence on surfaced metal wearout and hardness.

Introduction

Mechanisms of the mining equipment, which works in abrasive and shock wear conditions, premature brakes down and requires repairmen. That is why developing materials, that increase wear resistance of such mechanisms and details, and developing technologies of their recovering is the main objective. Most perspective for surfacing wearout details is automatic flux cored wire depositing. For these purpose, in Russia and foreign countries using, developing and manufacturing of specialized surfacing flux cored wires is in progress [1-16]. In balanced way of alloying response, deposited metal exhibit high level of hardness, abrasive and abrasive-shock wear resistance. Nowadays in Russia is widely used flux cored wire of the system Fe-C-Si-Mn-Cr-Ni-Mo DRATEC brand (Germany) type DT-SG 600 F and flux cored wires of ESAB brand types OK Tubrodur 15.52, OK Tubrodur 58 O/G M.

Materials and Technology

This work continue later researches in developing new flux cored wires chemical compositions for surfacing details of equipment that works in mining industry abrasive wear conditions [17-18], particularly studying the influence of tungsten and chromium appending in system Fe-C-Si-Mn-Mo-Ni-V-Co on wearout and hardness level of deposited layers.

Flux cored wire manufacturing was conducted on laboratory machine. Diameter of powder wire is 5mm, wire shell is made from metal ribbon of St3 grade. As a filler were used corresponding powder materials: iron powder brand PJV 1 by GOST 9849-86, ferrosilicon powder brand FS 75 by GOST 1415-93, powder of high carbon ferrochrome brand FH900A by GOST 4757-91, powder of carbon ferromanganese FMn 78(A) by GOST 4755-91, nickel powder PNK-1L5 by GOST 9722-97, ferromolybdenum powder brand FMo60 by GOST 4759-91, tungsten powder brand PVP-1 by TI 48-19-71-78, ferrovanadium powder brand FV50U 0,6 by GOST 27130-94, cobalt powder brand PK-1U2 by GOST 9721-79, As a carbon-contained component was used carbon-fluorine contain material with the following chemical composition, wt. %: Al₂O₃ = 21-46.23; F = 18-27; Na₂O = 8-15; K₂O = 0.4-6; CaO = 0.7-2.3; Si₂O = 0.5-2.48; Fe₂O₃ = 2.1-3.27; C = 12.5-30.2; MnO = 0.07-0.9; MgO = 0.06-0.9; S = 0.09-0.19; P = 0.1-0.18.

Surfacing under the AN-26S flux was conducted in 6 layers (to prevent mixing deposited metal with plates metal) on plates of the 09G2S grade steel, with the use of automatic welding machine ASAW-1250, under following welding conditions: I_w=450 A; U_a= 30 V; V_w=10 meter per hour.

After the plates were cut on appropriate samples for tests. Chemical composition of the researched samples was defined according to GOST 10543-98 by X-ray fluorescence method on XRF-1800 spectrometer and by atomic emission spectrometer DFS-71. Chemical composition of the surfaced by flux cored wires samples is shown in table 1. Hardness of the researched samples (table 2) was measured with the use of hardness meter MET-DU. Wear resistance (table 2) tests were conducted on machine 2070 SMT – 1. Every test took 6 hours with capacity 30 mA and rotation frequency 20 revolutions per minute. Studying the microstructure of the samples was conducted on optical microscope OLYMPUS GX-51 in the light field in the limits of range $\times 100$ -1000 magnification after etching samples by nital to determine the degree of influence changes in chemical composition on parameters of the microstructure. Size of former austenite grain was determined by GOST 5639-56 with $\times 100$ magnification. Martensite needles size was determined by GOST 8233-56 with $\times 1000$ magnification. Studying of longitudinal deposited metal samples on nonmetallic inclusions existence was conducted in according with GOST 1778-70 with $\times 100$ magnification.

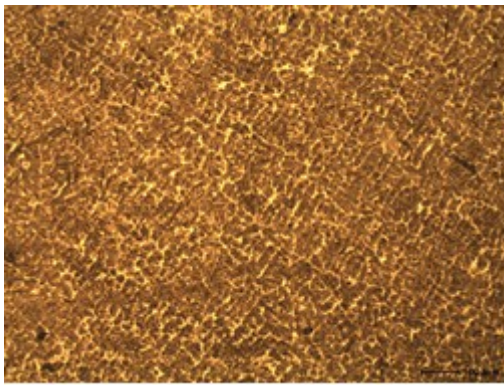
Results and Discussion

Table 1. Chemical composition of surfaced samples.

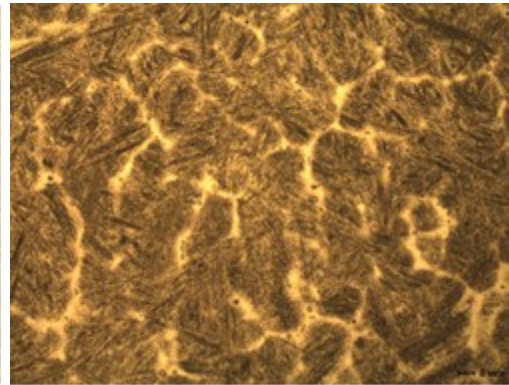
Sample number	Components weight content %													
	C	Si	Mn	Cr	Ni	Mo	V	Co	W	Al	Cu	Ti	S	P
1G9	0.19	0.77	0.61	4.17	0.34	0.38	0.02	0.05	0.002	0.01	0.07	0.02	0.054	0.024
2G10	0.19	0.63	0.65	4.06	0.3	0.38	0.03	0.06	0.001	0.01	0.08	0.03	0.056	0.019
3G11	0.2	0.59	0.61	4.12	0.3	0.38	0.02	0.12	0.001	0.02	0.06	0.04	0.049	0.019
4G12	0.2	0.64	0.6	4.03	0.3	0.39	0.03	0.2	0.002	0.01	0.08	0.03	0.058	0.021
5G13	0.2	0.59	0.56	0.01	0.3	0.33	0.01	0.05	7.74	0.01	0.1	0.02	0.072	0.017
6G14	0.2	0.55	0.49	0.01	0.26	0.34	0.01	0.07	7.42	0.02	0.09	0.02	0.048	0.014
7G15	0.2	0.58	0.52	0.01	0.28	0.34	0.01	0.07	7.55	0.05	0.09	0.03	0.038	0.014
8G16	0.21	0.55	0.52	0.01	0.27	0.35	0.02	0.06	7.65	0.05	0.08	0.02	0.036	0.017

Table 2. Hardness and wear intensity of surfaced samples.

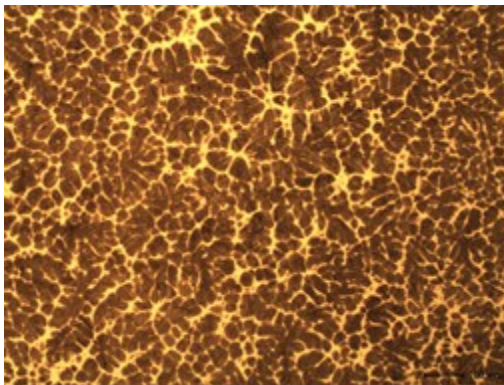
Sample number	Hardness, HRC	Wear intensity, grams per revolution
1G9	44.5	0.000071
2G10	43	0.000039
3G11	46	0.000044
4G12	30	0.000073
5G13	22	0.000206
6G14	25	0.000048
7G15	21	0.000039
8G16	25	0.000036



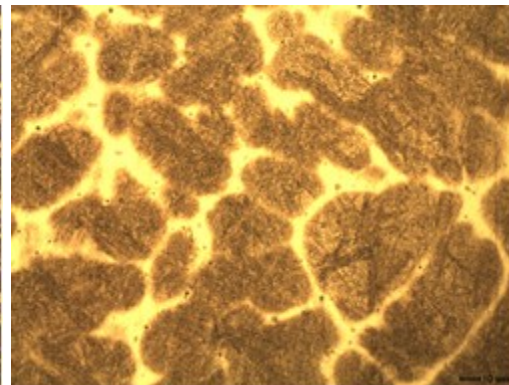
a



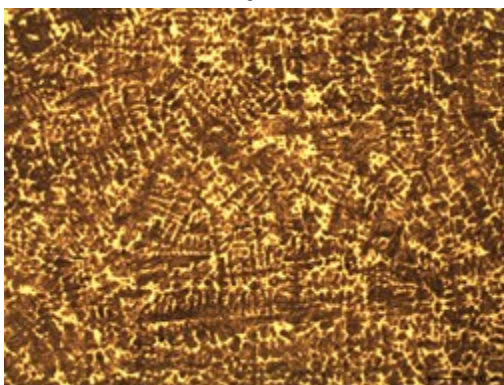
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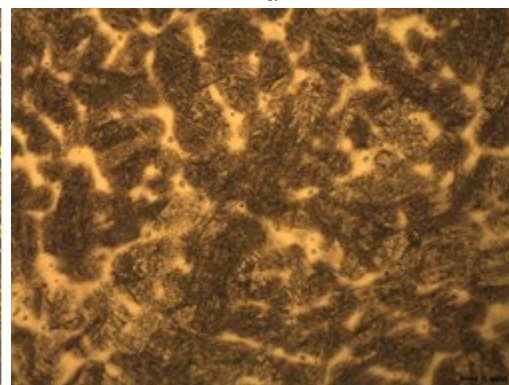
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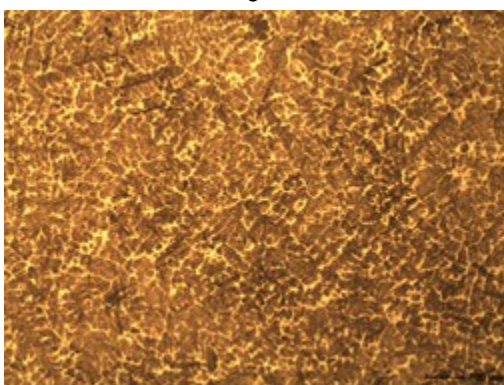
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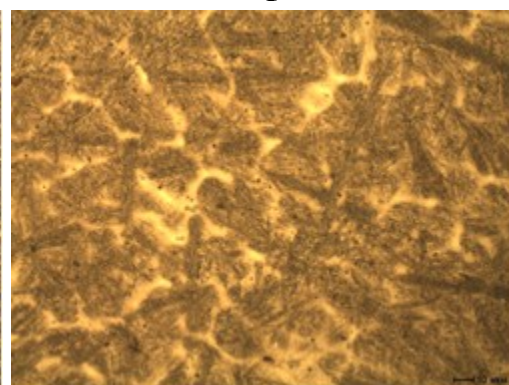
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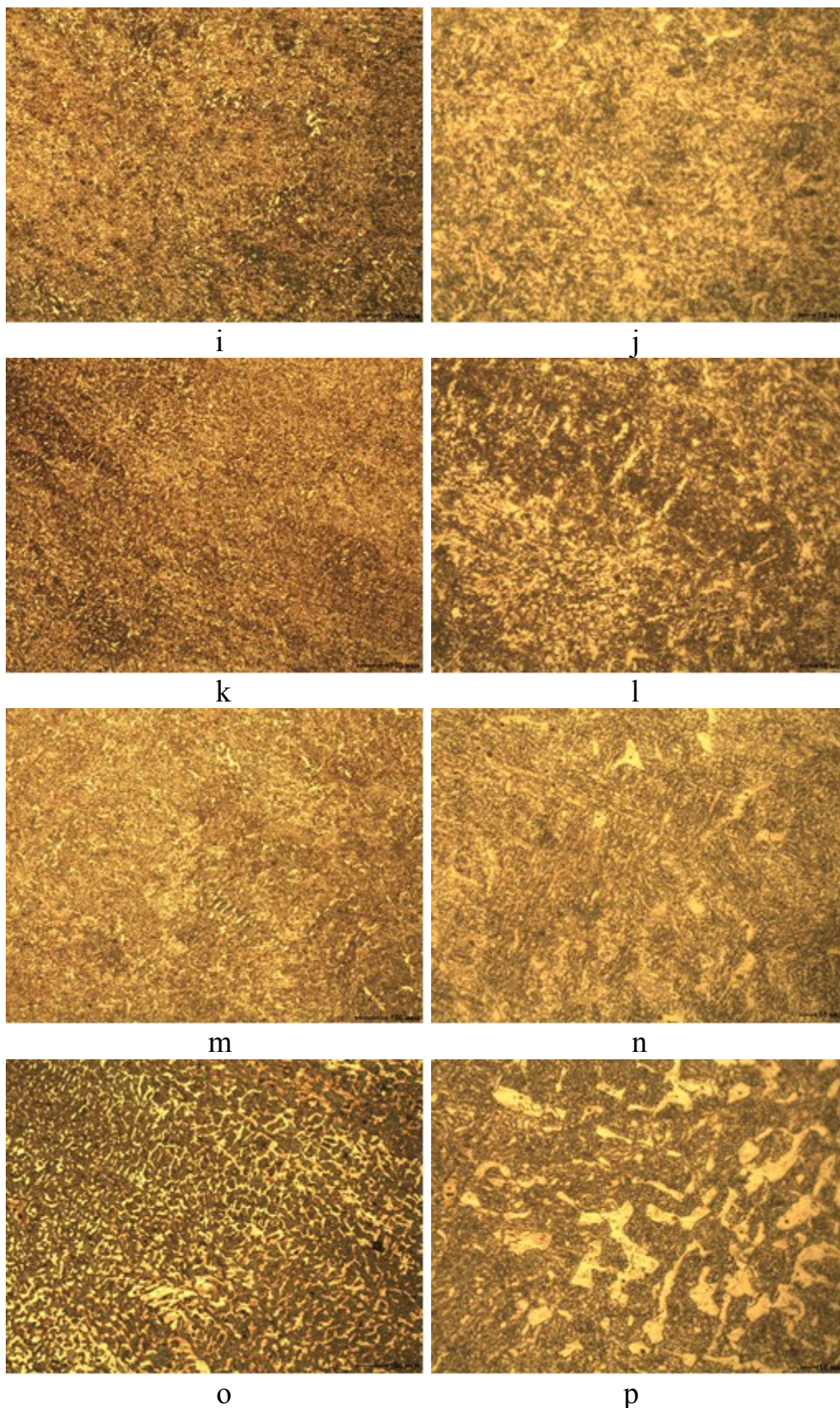
f



g



h



a, b - №1G9; c, d - №2G10; e, f - №3G11; g, h - №4G12; i, j - №5G13; k, l - 6G14; n, m - 7G15; o, p - 8G16.

Fig. 1. Microstructure of the researched samples, (a, c, e, g, i, k, m, o $\times 100$), (b, d, f, h, j, l, n, p $\times 500$).

The structure of №1G9 sample (Fig. 1 a, b) contain big-needled martensite (rating №7) in former austenite grains, in which borders is presented unlocked thin layers of something what is proposed to be comprise of δ -ferrite. Martensite needles size is in ranges of 3-14 mcm. Deposited layer could be characterized by surely dendritic structure signified. Austenite grain size conforms №6 and 7 in grain fineness scale compliance. All round the body and in borders of grains, dispersive inclusions are located.

The structure of №2G10 sample (Fig. 1 c, d) contain fine- and middle-needled martensite (rating №4, 5) in former austenite grains, in which borders is presented unlocked thin layers of something what is proposed to be comprise of δ -ferrite. Martensite needles size is in ranges of 3-6 mcm. Deposited layer could be characterized by surely dendritic structure signified. Austenite grain size conforms №6 and 7 in grain fineness scale compliance. All round the body and in borders of grains, dispersive inclusions are located.

The structure of №3G11 sample (Fig. 1 e, f) contain big-needled martensite (rating №7), which is formed inside clearly seen borders of former austenite grains. On the borders of former austenite, unlocked thin layers of something what is proposed to be comprise of δ -ferrite are located. Martensite needles size is in ranges of 8-12 mcm. Deposited layer could be characterized by surely dendritic structure signified. Austenite grain size conforms №6 and 7 in grain fineness scale compliance. All round the body and in borders of grains, dispersive inclusions are located.

The structure of №4G12 sample (Fig. 1 g, h) contain big-needled martensite (rating №7), in former austenite grains, in which borders is presented unlocked thin layers of something what is proposed to be comprise of δ -ferrite. Martensite needles size is in ranges of 3-11 mcm. Deposited layer could be characterized by surely dendritic structure signified. Austenite grain size conforms №6 and 7 in grain fineness scale compliance. All round the body and in borders of grains, dispersive inclusions are located.

The structure of №5G13 sample (Fig. 1 i, j) contain ferrite and perlite. Former austenite grain size conforms №5 and 6 in grain fineness scale compliance. All round grain body and ferrite, spot inclusions are located.

№6G14 sample (Fig. 1 k, l) have hypoppearlitic structure. Former austenite grain size conforms №6 in grain fineness scale compliance. All round grain body and ferrite, spot inclusions are located.

№7G15 sample (Fig. 1 m, n) have hypoppearlitic structure. Former austenite grain size conforms №5 in grain fineness scale compliance. All round grain body and ferrite, spot inclusions are located.

№8G16 sample (Fig. 1 o, p) have hypoppearlitic structure. Former austenite grain size conforms №6 in grain fineness scale compliance. All round grain body and ferrite, spot inclusions are located.

Chromium and tungsten concentration influence on samples hardness is shown on Figure 2 and 3. Despite appending tungsten in 5G13-8G16 samples chemical composition, their wear resistance is predetermined by the structure of deposited metal. Wear resistance of samples 1G9-4G12 is higher because they have martensite structure, and 5G13-8G16 samples have hypoppearlitic.

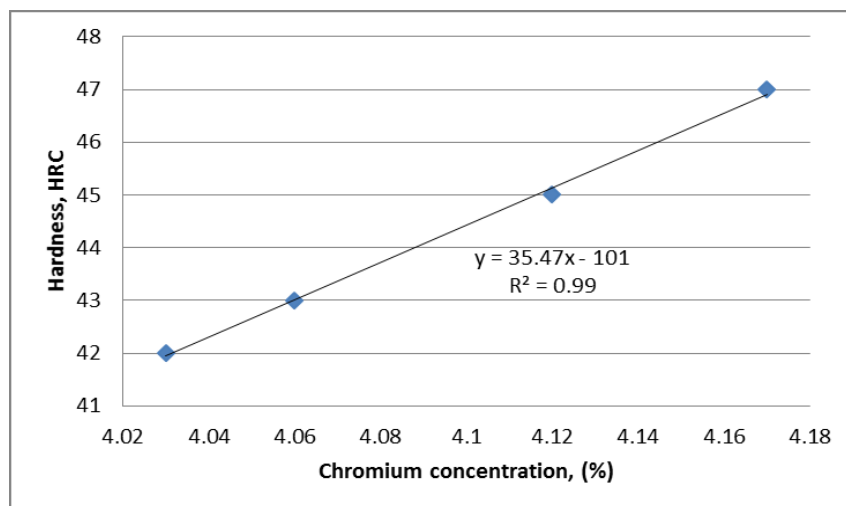


Fig. 2. 1G9-4G12 samples hardness.

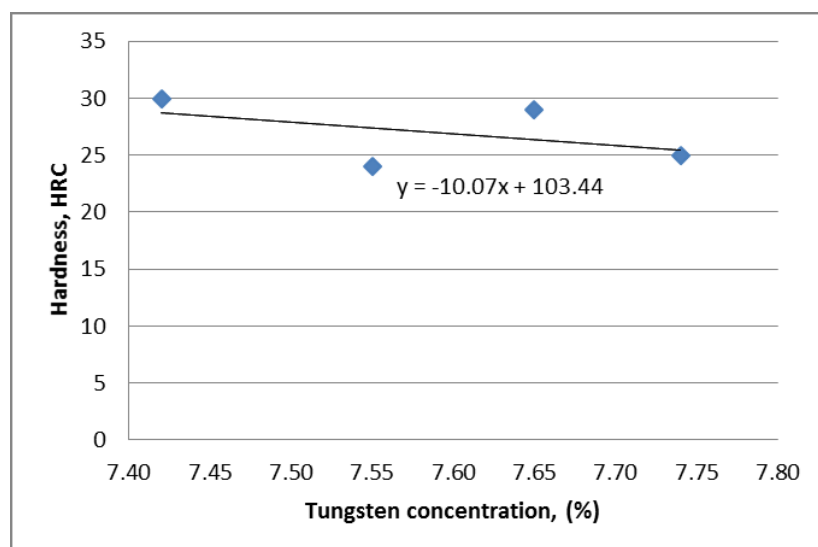


Fig. 3. 5G13-8G16 samples hardness.

According to the results of samples nonmetallic inclusions impurity research could be established the presence of nonmetallic inclusions, particularly nondeformable silicates and oxide spots (table 3).

Table 3. Nonmetallic inclusions and structure characteristics of the researched samples.

Sample number	The contamination of nonmetallic inclusions, rating		Size of austenite grain, rating	Size of martensite needles, mcm
	Non-deformable silicates (fragile)	Oxide spots		
1G9	1b, 2b, 3a	1a	6, 5	2-5
2G10	1b, 2b, 3a	1a, 2a	6	2-4
3G11	1b, 2b, 3a	1a	6	2-5
4G12	1b, 2b, 3a	1a	6	2-4
5G13	2b, 1b, 2a	1a	4, 5	-
6G14	2b, 1b, 2a	1a	4, 5	-
7G15	1b, 2b, 2a	1a	4	-
8G16	2b, 3b	1a	4	-

Conclusions

1. Appending chromium in the system Fe-C-Si-Mn-Mo-Ni-V-Co provides forming martensite structure in the deposited layer, with thin layers of δ -ferrite, supplies requiring hardness and wear resistance for the details which works in abrasive wear conditions.
2. Appending tungsten in the system Fe-C-Si-Mn-Mo-Ni-V-Co and almost absence of chromium in its charge, provides forming hypopearlitic structure in deposited layer with the tungsten carbide elements, while didn't supplying requiring hardness and wear resistance level.

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The Change in the Thin Structure and Mechanical Properties of Aluminum Alloys at Intensive Plastic Deformation

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Keywords: deformation, aluminum alloy, thin structure, strain curves.

Abstract. Electron microscopy and x-ray analysis and mechanical testing have been investigated the influence of severe plastic deformation on structure and mechanical properties of aluminum alloys. It is established that in the initial state in the alloy AMC has a high density of chaotically distributed dislocations with a density of $5-10 \cdot 10^9 \text{ cm}^{-2}$. It is shown that microdiffraction paintings in alloy AMC in the bulk of grains are observed uniformly distributed particles of the second phase. It is established that in the initial state in the alloy AMG6 there is a high density of chaotically distributed dislocations with a density of $2-6 \cdot 10^{10} \text{ cm}^{-2}$. Determined that after ECAP the dislocation structure of alloys AMG6, AMC and changes: formed dislocation networks inside the fragments of the dislocation is practically not observed. Determined that after ECAP-12 increase the tensile strength and yield strength of alloys AMG6 and AMC.

Introduction

As is known, improvement of mechanical characteristics of aluminum alloys by the application of traditional heat treatments is studied in detail and almost exhausted. The disadvantages of this method is limited scope (only wire), temperature of operation, as well as obtaining high strength, low ductility of the material [1-3]. Further progress in the solution of these tasks can be achieved by use of advanced deformation methods. Therefore, in recent years more and more attention of researchers aimed at the development and research of SMC and NC States obtained by intensive plastic deformation (IPD) techniques in inexpensive alloys such as aluminum alloys, which can significantly expand the class of structural materials by creating increased strength properties in them [4-6].

One of the known methods of IPD, allowing to provide high intensity and a more homogeneous strain state is equal channel angular pressing (ECAP). Today this method has been actively studied and used to obtain volume SMC materials [7]. Other methods for IPD: a comprehensive forging, deformation by torsion under high pressure, shear pressure, screw extrusion, low-temperature deformation and have a number of disadvantages, primarily related to the economic indicators of the processes. However, despite the significant number of experimental works, produced by IPD in aluminum alloys still have not found features of the fine structure of deformed aluminum alloys.

In connection with the foregoing, the aim of this work is to study the influence of IPD on the fine structure and mechanical properties of aluminum alloys.

Material and Experimental Procedure

As research material was selected as aluminum alloys AMG6, AMC and widely used in modern aviation, aircraft, shipbuilding, railway transport, road transport, construction, petroleum and chemical industries. Chemical composition of aluminum alloys: AMC – 96.35-99% Al; 1-1.5% Mn; 0.6% Si; 0.7% Fe; 0.05-0.2% of Cu; 0.1% Zn, AMG6-91.1-93% Al; 0.5-0.8% Mn; 0.4% Si; 0.4% Fe; 0.1% Cu; 0.2% Zn; 5.8-6.8% Mg. The study used an initial billet of aluminum alloy AMC and AMG6 12 mm in diameter and 50 mm long. The diameters of the channels of the extrusion were as

follows: the input (vertical) of 13 mm, output 12 mm. the Angle of intersection of channels pressing were respectively 90 and 120°. Used routes pressing: Sun (90° rotation) and C (rotated 180°). ECAP was performed using tooling that was installed on a hydraulic press brand KNWP 30M. Hydraulic press gives maximum pressure up to 40 MPa (30 tons). The pressed samples were produced with pressure up to 30 kg/cm², the pressing time up to 40 s on each pass.

Structural investigations were carried out by electron diffraction microscopy of thin foils in an electron microscope PHILIPS CM-12 operating magnification in the column of the microscope was chosen from 10,000 to 50,000 times. The phase composition was analyzed by indexing Mikroelektronics. For evaluation of mechanical properties tensile tests of samples of alloys.

Results and Discussion

The results of the study of the mechanical properties of aluminum alloys AMG6, AMC, and showed that as a result of equal channel angular pressing, the microhardness of alloy AMG6 increases almost 4 times in comparison with the initial state, the microhardness of alloy AMC increases, which is almost 4.5 times higher than in the initial state. After ECAP-12 mass loss is reduced to 5.4 and 5.6 mg, which shows an increase in wear resistance of aluminum alloys AMG6 and AMC 13-14 % [8].

Also the main objective of this work was to increase the tensile strength of aluminum alloys AMG6 and AMC, while maintaining or increasing ductility. Obtaining the alloy with such properties will allow in many cases to replace expensive aluminum alloys, type 1575. In this regard, were investigated mechanical properties of alloys AMG6 and AMC before and after ECAP. Figure 1 shows the deformation curves of alloys AMG6, AMC is in the initial state and after ECAP.

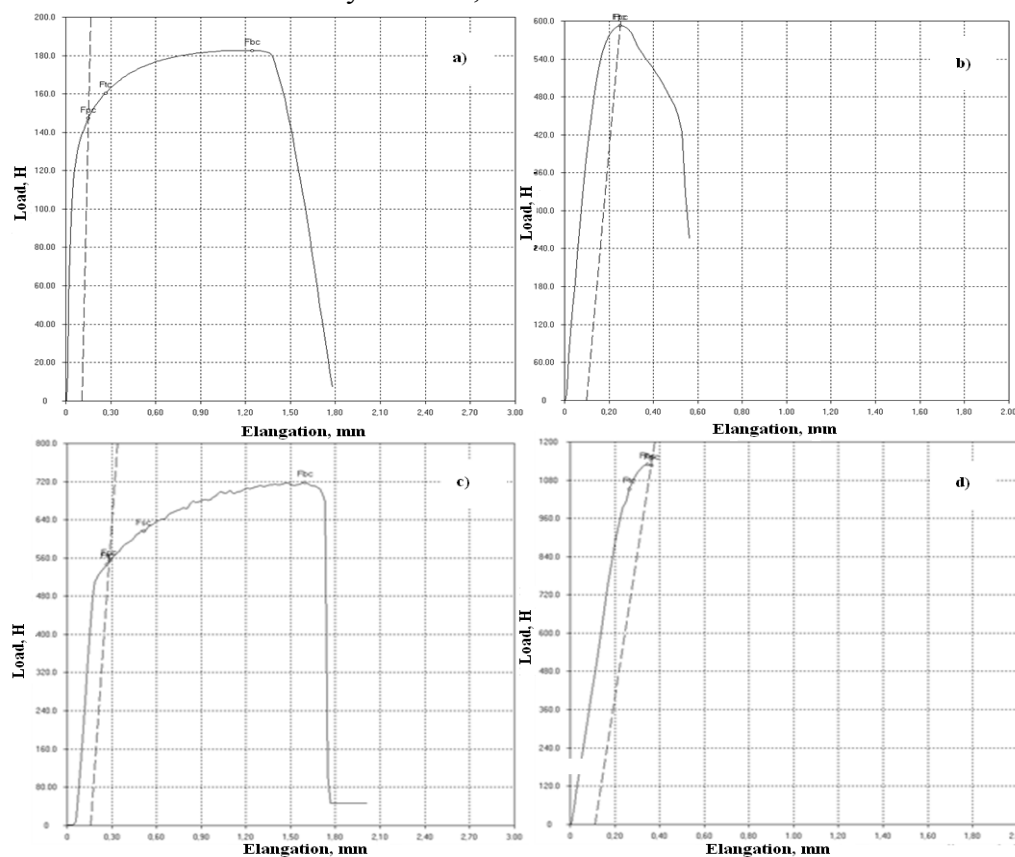


Fig. 1. Strain curves of alloy AMC (a,b) and AMG6 (c,d) in initial States (a,c) and after ECAP-12 (b,d).

Comparison of the curves indicates a significant difference of the strength characteristics of the original and deformed ECAP samples of alloys AMG6 and AMC. Other outstanding features are the length of the stage of strain hardening. From the curves it follows that the deformation hardening at the stage of plastic flow in the ECAP samples is smaller than in the original state.

Table 1 shows the mechanical properties of samples of alloy AMG6, AMC and before and after ECAP. After ECAP-12 alloy AMC increase its tensile strength of 91 MPa and 297 MPa, and a yield strength of between 75 MPa and 297 MPa. After ECAP-12 alloy AMG6 increase its tensile strength of 359 MPa and 565 MPa, the yield strength of 279 MPa and 565 MPa. While the values of the strength characteristics increase, the plasticity of alloys AMG6, AMC and falls in the transition to nanostructured state during ECAP. The value of elongation to failure of samples of alloys AMG6, AMC, and reduced to 40% and 20%, respectively.

Table 1. Mechanical properties of samples of AMC and AMG6 alloys.

Mechanical properties	AMC-initial	AMC-ECAP-12	AMG6-initial	AMG6- ECAP-12
Tensile strength, (MPa)	91	297	359	565
Yield strength, (MPa)	75	297	279	565
Elongation, (%)	5	3	10	8

The results of the study of the fine structure showed that in the initial state in the alloy AMC has a high density of chaotically distributed dislocations with a density of $5-10 \cdot 10^9 \text{ cm}^{-2}$ (Figure 2). In the bulk of grains are observed uniformly distributed particles of the second phase. On the microdiffraction detected reflexes from the face-centered cubic lattice phase with lattice parameter close to the lattice parameter of aluminum, as well as additional reflections from planes with interplanar distance $d_{hkl}=4.00 \text{ \AA}$, $d_{hkl}=2.87 \text{ \AA}$, $d_{hkl}=2.40 \text{ \AA}$, $d_{hkl}=2.24 \text{ \AA}$, $d_{hkl}=2.04 \text{ \AA}$.

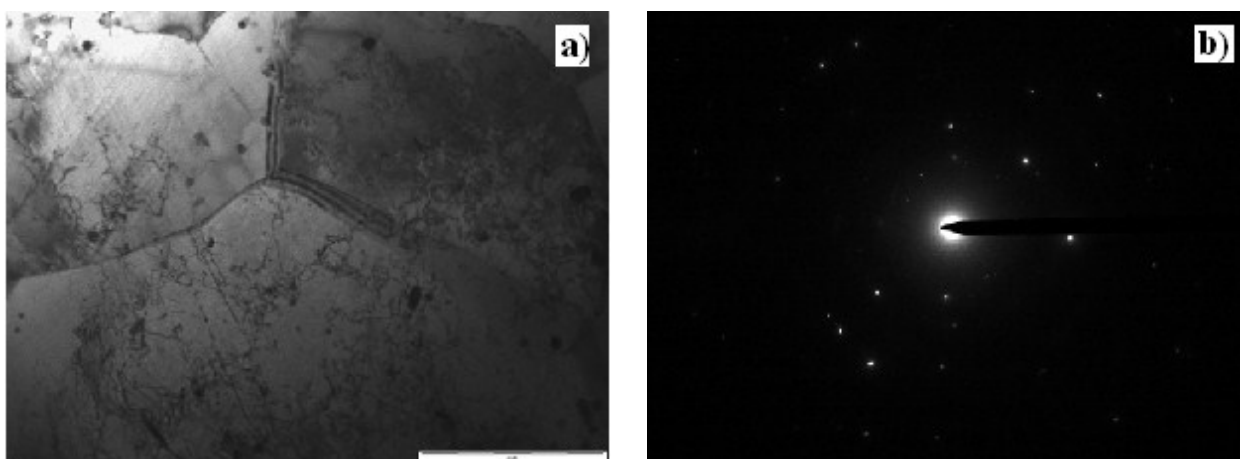


Fig. 2. Bright field image (a) structure and microdiffraction picture (b) plot of the deformed alloy AMC.

After ECAP observed splitting of unidirectional material into small fragments. On the microdiffraction detected reflexes from the face-centered cubic lattice phase with lattice parameter close to the lattice parameter of aluminum. Reflections from secondary phases are absent. The image of the structure of second phase particles is also observed. Dislocation structure changes: the dislocation grids are formed (indicated by arrows in figure 3), within the fragments of the dislocation is practically not observed. This structure is studied in detail in [9,10]. The size of the fragments (about $1 \mu\text{m}$) easy to estimate in figure 5. Darkfield analysis (figure 4) showed that most of disordering boundaries are small angle (about 2-5 degrees).

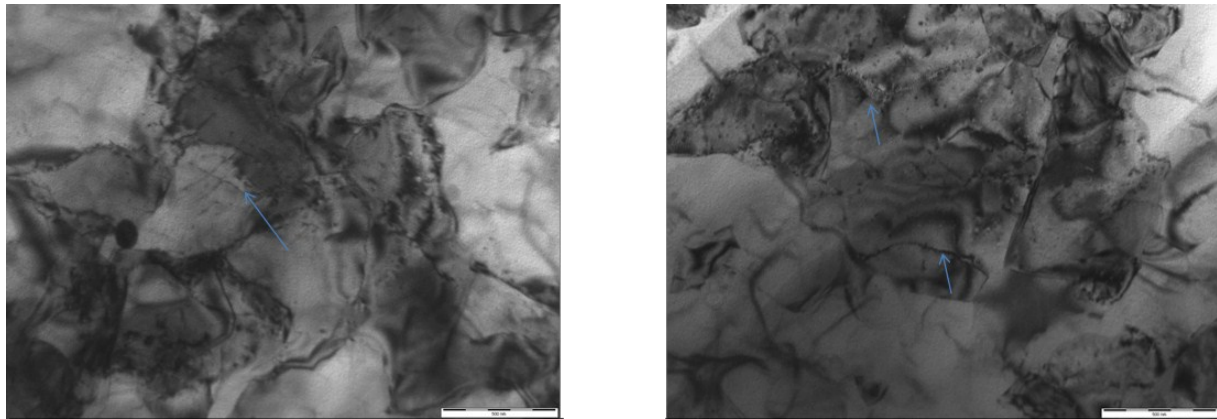


Fig. 3. Bright field image of the alloy structure AMC.

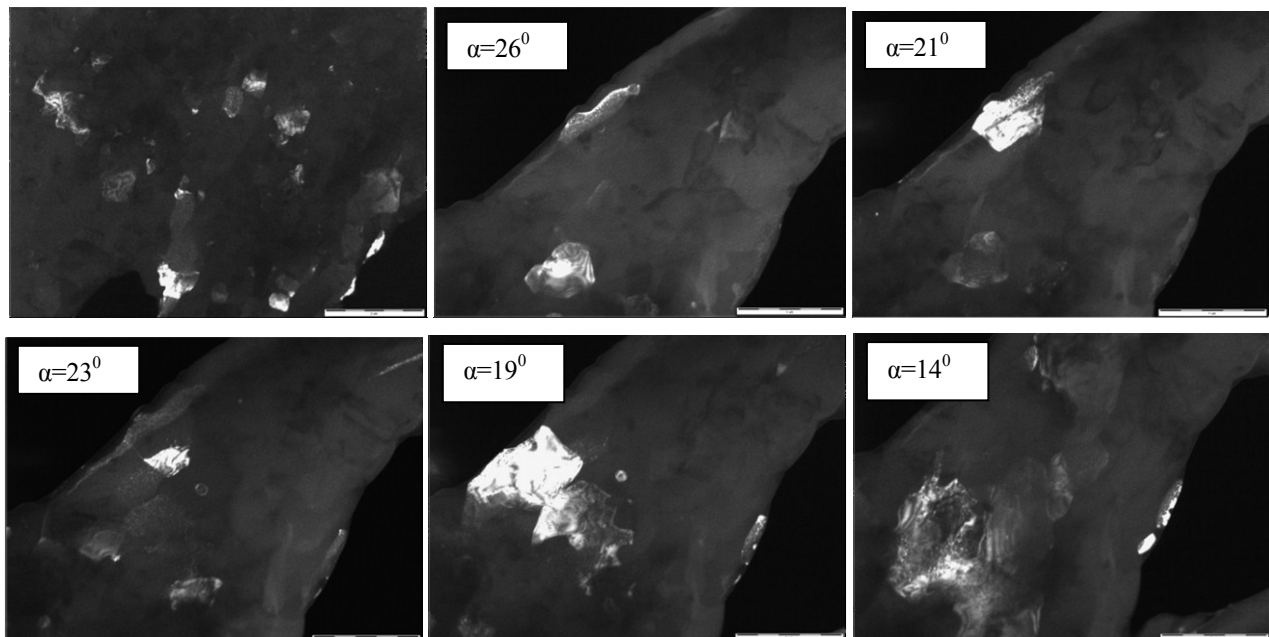


Fig. 4. Darkfield image of the structure of the AMC alloy after ECAP in matrix reflex (α – reading goniometer).

Method TEM was also studied fine structure of the wrought aluminum alloy system Al-Mg-Mn – alloy AMG6. Alloy AMG6 a combination of satisfactory strength, good ductility, corrosion resistance and good weldability. It defines the wide application of alloy AMG6 in shipbuilding, construction, rocketry and aviation.

Figure 5 shows a bright field image of the fine structure of alloy AMG6 and its microdiffraction. In the initial state in the alloy AMG6 there is a high density of chaotically distributed dislocations with a density of $2-6 \cdot 10^{10} \text{ cm}^{-2}$. In the bulk of grains are observed uniformly distributed particles of the second phase. On the microdiffraction detected reflexes from the face-centered cubic phase with lattice parameter close to the lattice parameter of aluminum, as well as additional reflections from planes with interplanar distance d_{hkl} of 2.57 Å, d_{hkl} of 2.35 Å, d_{hkl} of 2.15 Å, d_{hkl} of 2.09 Å, d_{hkl} of 1.68 Å.

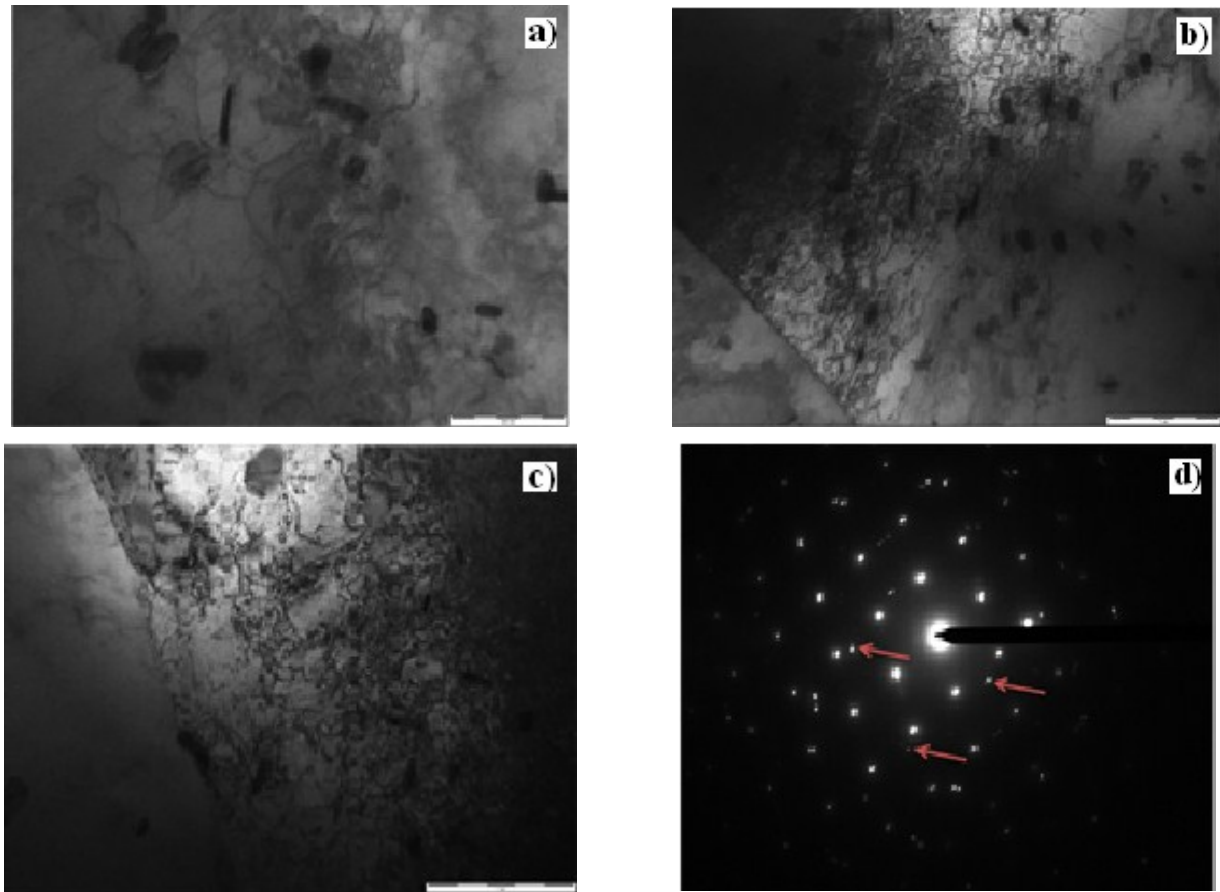


Fig. 5. Bright field images of the structure of alloy AMG6 - (a, b, c) and microdiffraction from the area of the sample (d). Plane (110) with parameters Al and a separate reflexes from the second phase (indicated by arrows).

After ECAP observed splitting of unidirectional material into small fragments. On the microdiffraction detected reflexes from the face-centered cubic phase with lattice parameter close to the lattice parameter of aluminum. There are separate reflexes from the second phase, however, the image structure of the second phase particles are almost not observed. Dislocation structure changes: the dislocation grids are formed (indicated by arrows in figure 6) inside the fragments of the dislocation is practically not observed. The fragment size of about 0.5 microns. Darkfield analysis (figure 7) showed that most of disordering boundaries are small angle (about 2-5 degrees).

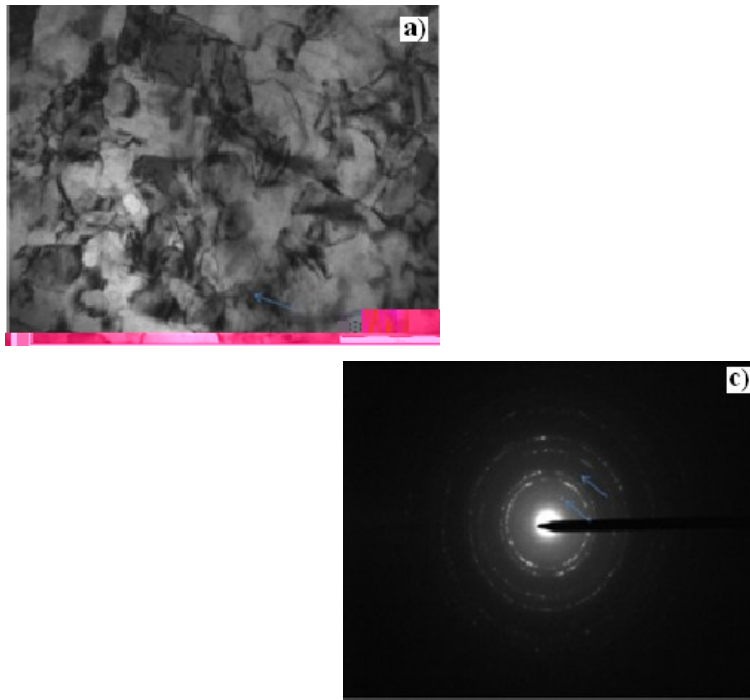


Fig. 6. Bright field images of alloy AMG6 - (a, b) and microdiffraction the plot of the sample - (c) of alloy AMG6 with parameters Al and a separate reflexes from the second phase (indicated by arrows).

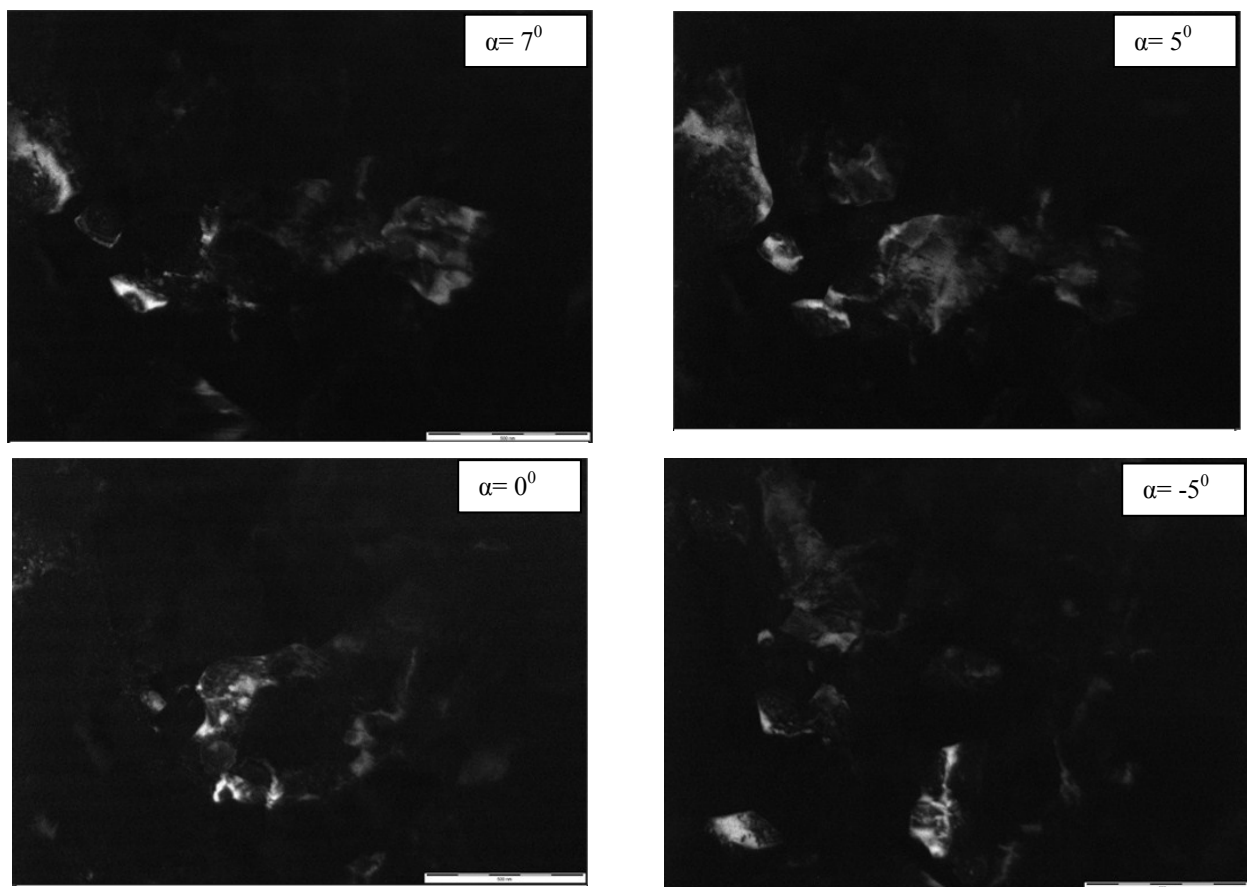


Fig. 7. Darkfield image of the structure of alloy AMG6 in matrix reflex (α – readings of the goniometer).

Conclusions

Thus, on the basis of the received results of research of influence of severe plastic deformation on structure and mechanical properties of aluminum alloys you can make the following conclusions:

- it is established that in the initial state in the alloy AMC has a high density of chaotically distributed dislocations with a density of $5-10 \cdot 10^9 \text{ cm}^{-2}$;

- it is shown that microdiffraction paintings in alloy AMC in the bulk of grains are observed uniformly distributed particles of the second phase;

- it is established that in the initial state in the alloy AMG6 there is a high density of chaotically distributed dislocations with a density of $2-6 \cdot 10^{10} \text{ cm}^{-2}$;

- determined that after ECAP the dislocation structure of alloys AMG6, AMC and changes: formed dislocation networks inside the fragments of the dislocation is practically not observed. The fragment size of about 0.5 microns. Darkfield analysis has shown that most boundaries are low angle disordering;

- determined that after ECAP-12 alloy AMC increase its tensile strength of 91 MPa and 297 MPa, and a yield strength of between 75 MPa and 297 MPa. After ECAP-12 alloy AMG6 increase its tensile strength of 359 MPa and 565 MPa, the yield strength of 279 MPa and 565 MPa;

- it is shown that ECAP AMC and AMG6 strength characteristics increase, alloys plasticity AMG6, AMC and falls in the transition to nanostructured state during ECAP. The value of elongation to failure of samples of alloys AMG6, AMC, and reduced to 40% and 20%, respectively. This work is executed at financial support of the Committee of Science Ministry of Education and Science on the theme "Obtaining SMC aluminum alloys with high physical and mechanical properties by severe plastic deformation".

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Welding Technology in Additive Manufacturing Processes of 3D Objects

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Keywords: welding, SLM-processes, 3D-printing process, properties.

Abstract. Using of welding technologies to produce metal volume objects allows considerable lowering of their manufacturing cost at simultaneous increase in productivity, compared to SLS- and SLM-processes. The most perspective welding technology of additive manufacturing of three-dimensional objects is plasma-arc technology with application of wires or powders. It allows creating at comparatively low heat input quality volumetric products with wall thickness from 3 to 50 mm from alloys based on Fe, Ni, Co, Cu, Ti, Al, as well as composite materials, containing refractory components.

1. Introduction

At present the additive manufacturing processes of 3D objects or rapid prototyping of complex-shaped volumetric products is considered to be the technology of XXI century, which will fundamentally change the structure of industrial production and economy, will enable computer-aided design of parts, flexible and fast manufacturing of different products, redistribution of production from large enterprises to small ones or manufacturing parts directly at user facility [1].

3D-printing is the technology of additive manufacturing. The process starts with obtaining virtual design data through computer modeling with application of CAD software. The machine for 3D-printing reads the data from CAD model file; CAM program modules are used to divide (cut) the part into layers, for each of which the tool movement trajectory is automatically generated, which can take into account a multitude of technological and geometrical factors. In total, all the programs for each layer form a control program, in keeping with which the CNC manipulator tool moves. The tool deposits successive layers of liquid, powder or sheet material, creating a physical model from a set of cross-sections.

These layers, corresponding to virtual cross-sections created in CAD model, are automatically connected to create the final shape.

There are different processes of additive manufacturing processes of 3D objects, but they are united by that the prototype is made by layer-by-layer deposition of material. The main advantage of rapid prototyping consists in that the prototype is created in one step, and the geometrical model of the part proper is used as input data for it. Therefore, there is no need for planning the sequence of technological processes, special equipment for material processing at each manufacturing stage, transportation from one machine tool to another one, etc.

Among the known and rather widely applied now processes of additive manufacturing processes of 3D objects we can note such techniques as stereolithography [2], fused deposition modeling (FDM) [3], and selective laser sintering (SLS) [4]. Such processes, with all their effectiveness, have one essential limitation: plastic is used in them as the main structural or binder material that considerably limits the range of manufactured items as to operating temperature, loads, mechanical strength and other values.

2. Enhancing the Capabilities of 3D-Printing

Enhancing the capabilities of 3D-printing requires availability of technologies of producing high-strength volumetric products from metals and alloys, including those of a high hardness. A number of US research centers (for instance, NASA's Langley Research center, Houston and Johnson Space centers, Hampton) perform development of electron beam process of manufacturing freeform metal items (EBF₃) [5]. Here, the electron beam is used as the energy source for melting the fed wire in vacuum. This procedure was demonstrated on aluminium and titanium alloys, which are of interest for aerospace applications [6]. In our opinion, it can be also extended to Ni- and Fe-based alloys. However, application of this process is limited by the need to apply expensive and complex vacuum equipment.

Table 1. Comparison of the most widely accepted SLM technologies of 3D-printing of metallic products with new plasma-arc melting technology based on plasma-arc welding.

Characteristics	Technology of 3D-printing of metallic products			
	SLM			PAM
	LENS company	POM company	AeroMet company	
Used equipment characteristic	Nd:YAG-laser	CO ₂ -laser	CO ₂ -laser	Based on welding
Power, kW	1	2	4	2-20
Productivity by incremental metal, cm ³ /h	8	8	160	>1000–15000
Capability of processing along axes (degrees of freedom), number of axes	3	3	3	4-5
Type of material used for 3D-printing	Metal powder			Powders of metals, alloys, composite materials, powder mixtures. Solid and flux-cored wire
Material utilization factor, %	About 40	About 40	About 70	More than 90
Application fields	Manufacture and repair of small-sized expensive complex-shaped parts			Manufacture and repair of medium- and large-sized expensive complex-shaped parts for various applications
Tentative cost of main equipment units, USD per 1 kW of power	Laser cost: 80,000–120,000			Welding equipment cost: 1000–5000

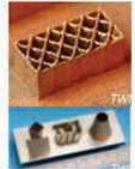
One of the promising technologies of producing high-strength metallic volumetric products is selective laser melting (SLM), ensuring item formation by fusing powders of different metals and alloys by the laser beam [7]. This technology enables producing complex metallic volumetric products with a high degree of detailing of their elements and high density (up to 99 %), as well as high dimensional accuracy ($\pm 50 \mu\text{m}$).

On the other hand, with all its effectiveness and flexibility SLM process also has a number of limitations, narrowing its application (Table 1):

- need to apply expensive and energy-consuming equipment with a high service cost that results in high cost of 3D-printing process and leads to a high cost of manufactured products;
- relatively low productivity of 3D-printing (usually not more than 10 cm³/h of incremental metal for the most widely applied machines);

- material limitations, namely for SLM expensive powders with strict requirements on granulometric and chemical composition and other characteristics are used;
- insufficiently high strength characteristics of manufactured items.

Table 2. Main technologies of additive manufacturing with application of local melting.

	Spraying				Powder/Mould		
	Plasma (wire)	EB (wire)	Laser (wire)	Laser (powder)	Laser (powder)	EB (powder)	Binder (powder)
Process schematic							
Illustration							
Description	Free spraying of wire using arc plasma	Spraying of wire molten by laser or electron beam in a chamber		Deposition of powder molten by laser in a chamber		Laser or electron beam selective melting in mould located in a chamber	Powder/binder system, requiring melting in the downhand position
Application	 High-rate melting of material and its spraying technology allow growing almost precise items			 Precise and almost finished sprayed parts	 New generation prismatic components with highly complex geometry		 Complex-shaped parts with inner cavities for cars

In view of the above-said, consideration of welding technologies for 3D-printing of complex-shaped metal products is urgent, as welding, at higher productivity, also allows realization of the principle of additive manufacturing, namely layer-by-layer formation of 3D structures. Moreover, welding technologies were developed long before appearance of 3D-printing, and are much more mature and less costly. Therefore, it is highly relevant to use welding processes in development of cost-effective method of manufacturing maximum dense metallic volumetric parts and tools [8].

Work [9] gives the following chronology of the attempts to apply welding technologies for fabrication of 3D structures of a complex-shape:

- in 1926 Baker patented «application of electric arc as the heat source to produce bulk objects by spraying molten metal into the deposited layers»;
- in 1971 Yujie patented manufacture of a high-pressure vessel, using submerged-arc welding, electroslag technology and TIG welding to produce items with functionally-gradient walls;
- in 1983 Kussmaul used shape welding to manufacture large-sized products from high-strength steel (20MnMoNi5) of 79 t weight;
- in 1993 Prinz and Weiss patented a combined technology of incrementing material, using welding with milling in CNC machine tools (shape deposition manufacturing — SDM);
- in 1994–1999 the Cranfield University developed shaped metal deposition (SMD) technology for manufacturing engine shells for Rolls Royce corporation (Great Britain).

There are also data on attempts made in the 1960s in Germany to create 3D metal structures, using shape welding. Based on this process, such companies as Krupp and Thyssen organized manufacturing of large-sized parts of a simple geometry, for instance, high-pressure vessels of up to 500 t weight [10]. Successful attempts of arc welding application for manufacturing large-sized metal structures and products from austenitic steels were made by the Babcock & Wilcox Company (USA) in the form of development of technology named Shape Melting [11]. As was already noted, Rolls Royce Corporation is pursuing arc-welding application as the technology providing high forming productivity and lowering the level of wastes that may be generated at traditional processing during fabrication of items from expensive alloys [12]. At present, this corporation is

successfully introducing this technology for manufacturing different aircraft parts from expensive Ni- and Ti-base alloys.



Figure 1. Complexes for realization of additive welding technologies [9]: a – five-axes system with CNC milling and WAAM technology; b – WAAM process without machining realized using an anthropomorphic laser.

In addition to the above examples, research work on 3D arc welding is conducted in University of Nottingham (Great Britain), Wollongong University (Australia) and Southern Methodist University in Dallas (Texas, USA) [13]. Research teams from Indian Institute of Technology (Mumbai), Fraunhofer Institute of Production Technology, and Automation presented their conceptual ideas of combining a welding operation with milling. Characteristic defects in volumetric product forming by welding methods were also studied and ways to eliminate them were developed [14]. Need to monitor the temperature of incremental layers was also studied. Special attention was given to creation of products from titanium [15] and nickel [16] alloys for aerospace industry applications. Overall, main welding and related technologies for additive manufacturing can be presented in the form of Table 2 [17].

Paton Welding Institute (PWI) work also confirmed the fundamental possibility of forming large-sized volumetric structures by arc welding. One of the striking examples can be creation of 3D welded sculptures and pictures from titanium alloy by G. Dochkin using his own unique method developed in mid-1970s [18]. Individual production problems were also solved, related to manufacture of unique products for defense industry.

The above studies were mostly focused on applicability for 3D-printing of such welding processes as gas metal arc welding (GMAW) and gas tungsten arc welding (GTAW). These processes ensure good metallurgical adhesion, as well as protection of weld pool and incremental layers of products from oxide formation. However, these processes with their availability also have such disadvantages as considerable size of HAZ and quite large dimensions of incremental layer that leads to generation of undesirable temperature gradients and residual stress accumulation. Moreover, mostly standard welding wire is used as consumable material to form volumetric products that limits the chemical composition and properties of these items.

The above drawbacks can be eliminated by application of microplasma or plasma-arc melting, as well as microplasma or plasma-arc welding. These welding technologies are capable of providing a new level of 3D-printing, both compared to arc welding processes (GMAW, GTAW), and compared SLM, in view of the following advantages:

- temperature in plasma arc flame can reach 30,000°C, that is essentially higher than in standard electric arc. Therefore, plasma can melt practically any refractory material for layer-by-layer incremental manufacturing of volumetric products;
- minimum heating of earlier deposited layers at product formation, less than 5 % heat penetration into base metal;
- absence of metal spatter when building-up the layers, their extremely low mixing;

- possibility of adjusting in a broad range the thickness (0.5–5.0 mm) and width (1.5–50.0 mm) of deposited metal layer in additive building-up of volumetric products;
- ability to adjust the composition of gas atmosphere (reducing, inert, oxidizing) during additive building-up of layers in product forming;
- high cost-effectiveness and productivity of the process (by 2–3 times and more);
- applicability of a wide range of consumable materials (powders of metals, alloys, composite materials, powder mixtures, solid and flux-cored wires), including those from Fe, Ni, Co, Cu, Ti, Al alloys, as well as composite materials, containing refractory components (carbides, borides, and other for instance, WC, Cr₃C₂, TiC, TiB₂), refractory materials, composite materials with refractory components;
- possibility of changing metal composition during item formation, producing items with graded structure.

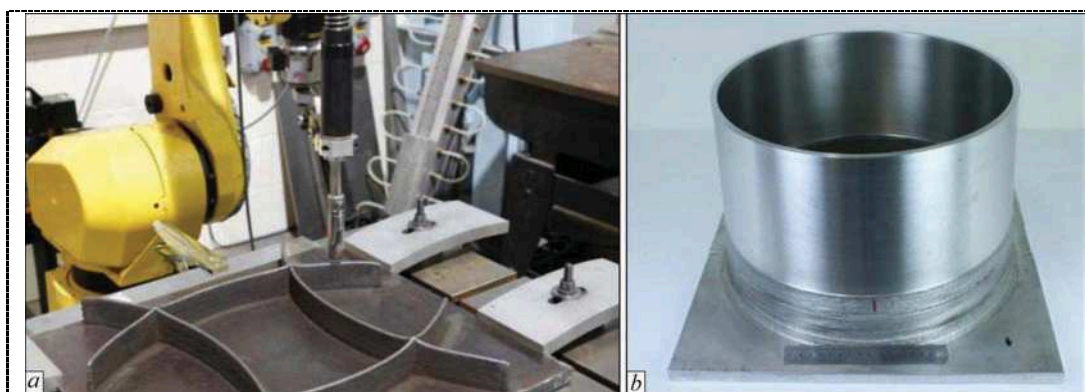


Figure 2. Examples of volumetric products from carbon steel S355 manufactured by WAAM technology [9]: a – panels with crossing stiffeners without machining; b – cylinder machined after manufacturing.

At present, advanced research institutes and industrial corporations of economically developed countries pursue applied research on development of plasma-arc melting and welding technologies, as well as other welding technologies of 3D-printing of metal products. Let us consider some characteristic examples of such studies.

Cranfield University developed different complexes for realization of additive welding technologies — both with machining during layer building-up and without it (Figure 1) [9]. Manufacturing volumetric products in these complexes is based on wire-arc additive manufacturing (WAAM) technology (Figure 2). Products are made from various materials, for instance, carbon steel, titanium and aluminium alloys, etc. Here, both standard (continuous or pulsed) and constricted electric arc, i.e. plasma, can be used (Figure 3). One of the promising directions of WAAM technology application is fabrication of large-sized honeycomb structures (Figure 4).

Southern Methodist University in Dallas studied the variants of laser, arc and plasma-arc technologies of manufacturing 3D objects with simultaneous or finish machining (CNC milling) [13]. Respective technological complexes were developed, and a number of engineering solutions were proposed, allowing manufacture of both relatively simple and quite complicated products (Figure 5). In particular, investigations were conducted in the field of microplasma powder building-up of volumetric products (Figure 6). Fundamental possibility of producing gradient composite structures by such a method was demonstrated. Analysis of the features and prospects for the above welding technologies showed that for the most high-efficient and cost-effective manufacturing of quality volumetric products with wall thickness of 3 to 50 mm, application of plasma-arc powder deposition of layers of 0.5–5.0 mm height in one pass is the optimum variant.



Figure 3. Shell casing of 160 mm diameter and 800 mm length with 8 to 18 mm wall thickness of 32 kg weight from high-strength steel made by WAAM technology (pulsed MIG welding) with productivity of 4 kg/h before (a) and after (b) machining [9].



Figure 4. Aluminium honeycomb beam structure of 3 m length with higher rigidity [9].



Figure 5. Process of manufacturing blade turbine by combination of arc surfacing and CNC milling: a – CAD model; b – start of GTA surfacing; c – end of GTA surfacing; d – blank after CNC milling [13].

In addition to machines and mechanisms, welding technologies of 3D-printing also allow creation of building structures. Development of new MX3D technology can be an example [19]. MX3D project was created by JORIS LAARMAN LAB in cooperation with ACOTECH and HAL (Holland). New technology has enormous potential, as it allows quickly creating complex metal constructions without erection of any accompanying supporting structures, for instance, scaffolding

or intermediate supports. During 3D welding, the man or welding robot creates his own support and moves ahead on the constructed structure. It speeds up and simplifies construction. Moreover, MX3D can be fully robotized, and can operate round the clock.

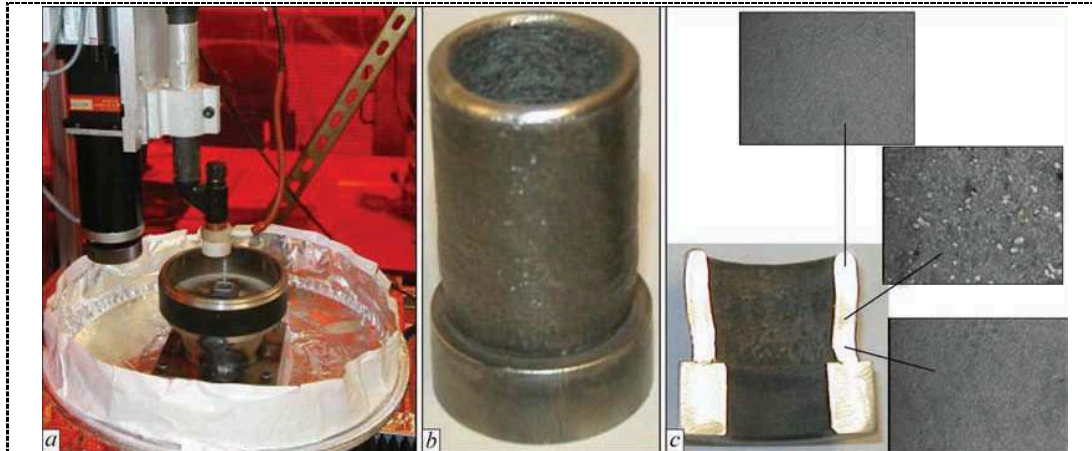


Figure 6. Microplasma powder growing of volumetric products [13]: a – laboratory unit; b – hollow cylinder of uniform structure (tool steel H-13); c – creation of gradient composite H-13 + tungsten carbide structure.

3. Equipment Based on Plasma-Arc Welding

Generalization of data, given and analyzed in publications, allows suggesting the following approach to fabrication of volumetric products, using welding technologies: computer modeling of the item, its manufacture by plasma-arc technology with application of wire or powder under the conditions of monitoring temperature and forming, machining of minimum required number of item sections. To realize such an approach, it is rational to separately apply the system for plasma-arc 3D-printing (Figure 7), and to perform machining in CNC milting and/or lathe machine tool. We believe it is not rational to combine the processes of 3D-printing and finish machining within one multi-axis all-purpose unit.

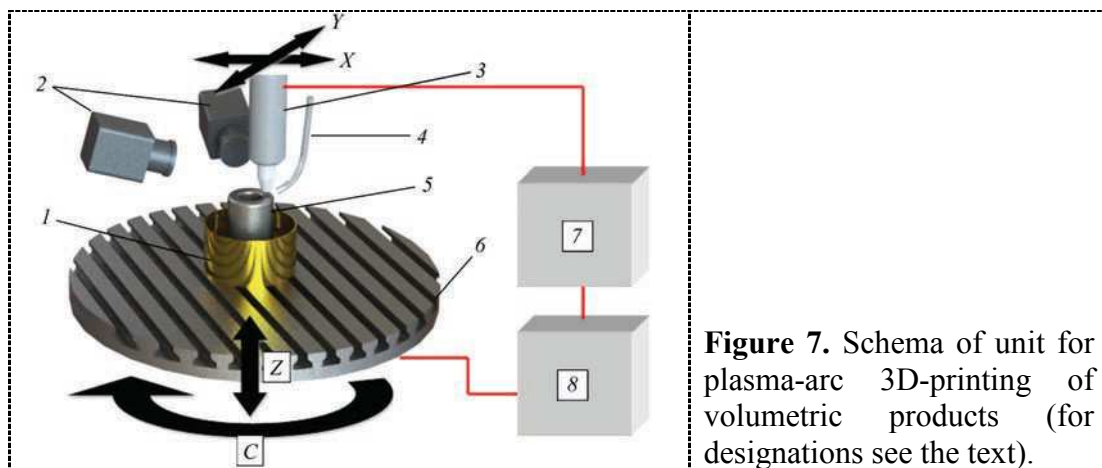


Figure 7. Schema of unit for plasma-arc 3D-printing of volumetric products (for designations see the text).

After development of a computer model of the grown part (using CAD system, for instance Solid-works), it is automatically (using CAM system, for instance Lazy CAM, Art CAM) separated into layers with generation of control programs for each of the layers, which are loaded into the system of CNC complex (see Figure 7). Growing of part 5 is performed in shielded (for capturing welding spatter and unused powder) zone 1, located on rotating table 6, with the capability of performing rotation C and step-by-step vertical displacement Z . Transferred arc plasmatron 3 is mounted on the carriage of xy -manipulator, providing precise movement along coordinates X and Y . Filler wire or powder is fed into the torch 3 coverage area using system 4. Process monitoring is performed by system 2, including two CCD-cameras, located at 90° angle to each other, and

thermal imager. Here, not only forming of part 5, but also its temperature state, is monitored. Systems 4 and 2 are also located on the carriage. Plasmatron 3 is powered by source 7, and 3D-printing process is controlled by CNC system 8. After growing part 5, it can be transferred to machining station, can be treated directly in the system, in which it was created, or left untreated, if required quality of its surfaces has been achieved.

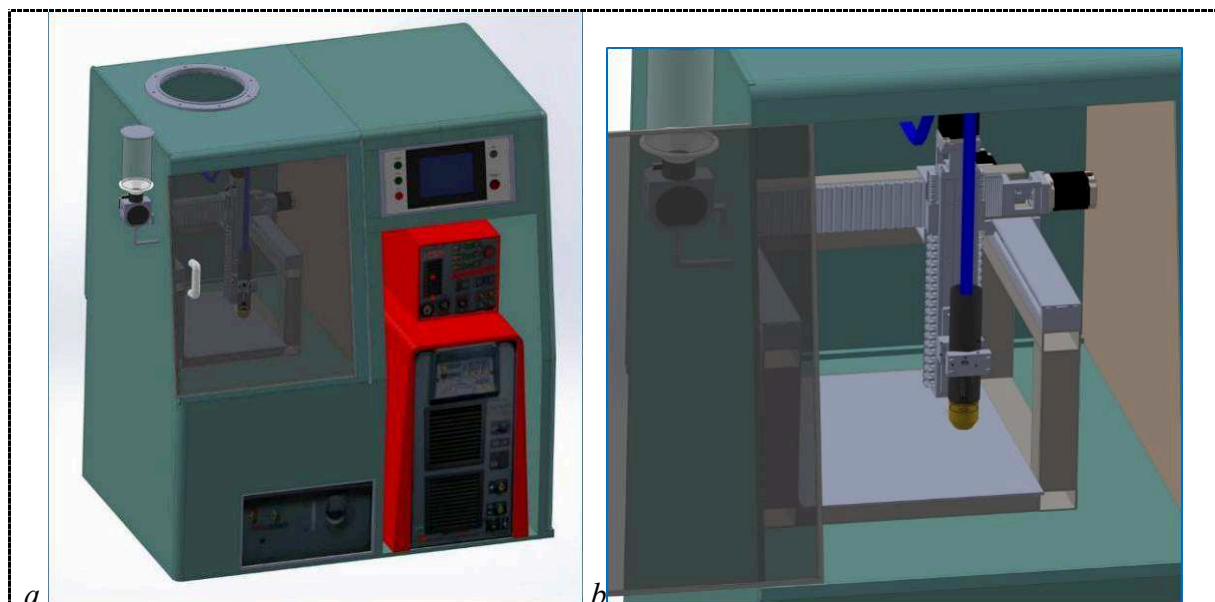


Figure 8. PWI-designed complex for manufacturing 3D objects based on technologies of plasma-arc cladding: a – general view; b – three-coordinate manipulator with plasmatron.

At present, PWI performs manufacturing of equipment for creation of 3D objects based on plasma-arc welding and surfacing. For this purpose, a complex was designed on the basis of three-coordinate manipulator, fitted with plasmatron with power source, filler wire feeder and CNC system (Figure 8, 9). The complex is controlled using a common controller with the capability of exchange of data on 3D-printing modes and of control commands with the computer. Performed work allowed development of a number of unified systems of different dimensions for 3D-printing of metallic volumetric parts of a complex shape, based on plasma-arc technologies.

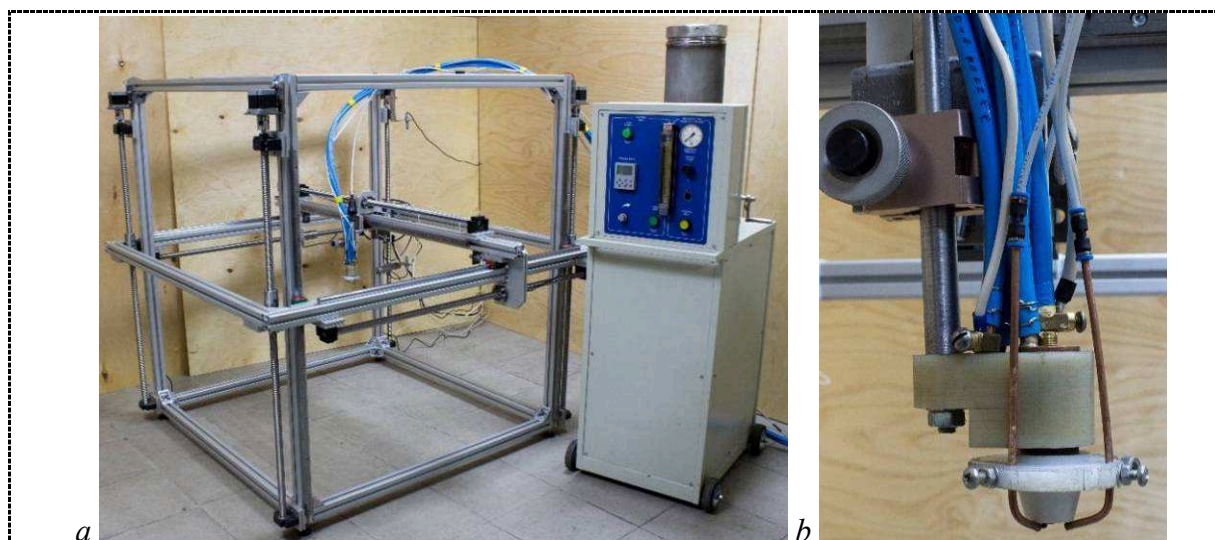


Figure 9. PWI-fabrication complex PLAZER 3D PW for manufacturing 3D objects based on technologies of plasma-arc cladding: a – manipulator with feeder and plasmatron; b – plasmatron.

Conclusions

1. Using of welding technologies to produce metal volume objects allows considerable lowering of their manufacturing cost at simultaneous increase in productivity, compared to SLS- and SLM-processes.
2. The most perspective welding technology of additive manufacturing of three-dimensional objects is plasma-arc technology with application of wires or powders. It allows creating at comparatively low heat input quality volumetric products with wall thickness from 3 to 50 mm from alloys based on Fe, Ni, Co, Cu, Ti, Al, as well as composite materials, containing refractory components.
3. Using of welding technologies allows producing both comparatively small and long items, not needing finish machining (for instance, growing stiffeners on large-sized panels, creating honeycomb panels, building structures, etc.).
4. Combination of welding technologies additive manufacturing of three-dimensional objects with concurrent or finish machining allows manufacturing finished metal items of complex profile.

Acknowledgements

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Aluminosilicate Microsphere Synthesis in Plasma Flow

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Keywords: microspheres, electropasma stand, porous particles, particle velocity, heating and melting of particles, physicochemical research.

Abstract. The paper presents theoretical and experimental results on the synthesis of ash-based microspheres using the high-temperature plasma treatment. The dynamic motion of particles and their melting are considered depending on the initial porosity of the raw material modified by 3200K plasma treatment. Physicochemical studies relate to the original raw material and microspheres produced therefrom.

Introduction

The morphology of powder particles subjected to thermal treatment directly depends on properties of raw materials such as the particle size distribution and chemical composition and electrophysical parameters of a heating source [1, 2]. Generally, conventional heating sources for spheroidization or synthesis of powder microspheres do not allow utilizing refractory raw materials with melt temperature not higher 1700 °C. The field of application of refractory raw materials can be widened with the use of plasma energy. A highly concentrated plasma energy provides a high temperature which facilitates the formation of a homogeneous structure inside the powder particles, thereby increasing their physical-and-mechanical properties [3, 4]. The plasma treatment of porous particles is a promising trend in the area of microsphere synthesis [5-7].

This paper mainly focuses on the experimental and theoretical research carried out into the use of porous aluminosilicate particles for the microsphere formation produced by the plasma flow. The process of microsphere synthesis can be divided into the following stages.

1. Heating of porous particles from initial to melt temperature. At this stage, air containing in pores expands and releases from the porous surface of a particle.
2. The formation of the primary external layer of a porous particle. During heating up to the melt temperature, a liquid layer of agglomerated particles forms on the surface of a porous particle. This layer encapsulates a certain air mass which is smaller than the original. Encapsulated air unifies into a hollow microsphere.
3. The formation of the external layer. The material inside the primary particle melts and precipitates on the internal surface of the liquid layer due to the air pressure and surface tension.
4. The layer amorphization during the particle cooling after their leaving the active zone of the high-temperature plasma (HTP) flow.

Materials and Methods

Bottom ash generated by GRES-1 power plant (Ekibastuz, Russia) was used in our experiment as a raw material. Table 1 contains the averaged chemical composition of bottom ash and maximum and minimum oxide contents.

Table 1. Chemical composition of bottom ash from GRES-1 power plant (Ekibastuz, Russia).

Raw material		Oxide content, wt. %							
		SiO ₂	Al ₂ O ₃	Fe ₂ O ₃	CaO	MgO	Na ₂ O	K ₂ O	Ignition loss
Ekibastuz	max	65.4	29.7	25.3	9.2	3.4	3.3	0.7	10.8
bottom ash	min	46.3	21.2	4.1	0.4	0.49	0.6	0.4	3.2
	averaged	55.85	25.45	5.65	1.17	0.8	1.95	0.42	5.1

Modifications of the compound composition induced by heating the raw material were traced on phase diagrams. The plasma-treated particles were considered in terms of the formation of silicate melt. For this consideration, a triangular phase diagram was constructed for a tertiary system (S–CAS2–A3S2) as shown in Fig. 1, with conversion factors introduced by the rule that low-melting oxides reduce the melt temperature of the material according to their equivalent weights. The course of the crystallization was found for aluminosilicate particles (Fig. 1), which locates in S–CAS2–A3S2 elementary triangle.

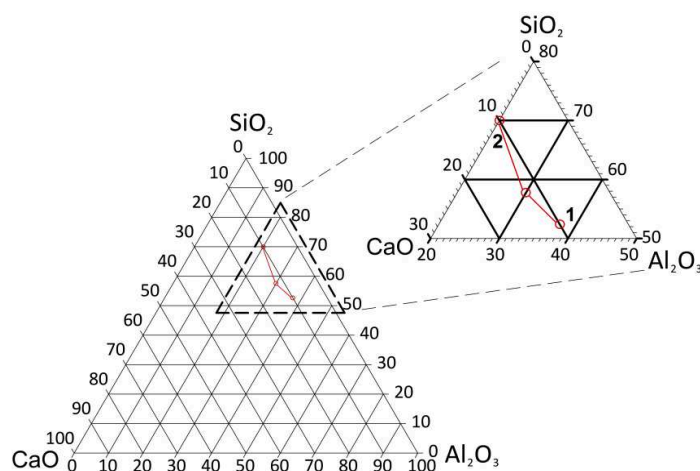


Fig. 1. Ternary phase diagram for crystallization course of bottom ash in S–CAS2–A3S2 elementary triangle.

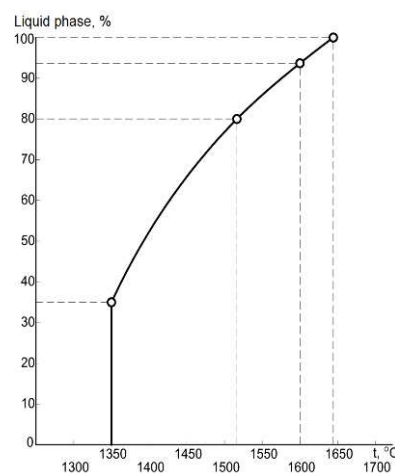


Fig. 2. Melting curve with regard to chemical composition of bottom ash.

Figure 1 presents a ternary phase diagram for Ekibastuz bottom ash, which locates in the section adjacent to SiO₂–Al₂O₃ side in the field of mullite crystallization. In other words, the formation of mullite crystals occurs during the melt cooling at 1640 °C initial temperature (Fig. 1, point 1). At temperature of 1540 °C, the simultaneous formation of mullite and anorthite crystals is observed. At the point of tertiary eutectic, the formation of quartz compound crystals occurs, and crystallization at 1345 °C is completed. Thus, at heating aluminosilicate particles up to 1345 °C, a liquid phase forms due to melting of tertiary eutectic. Further temperature growth enables melting of quartz compounds and gradual melting of aluminosilicate crystals. At point 2, all crystals melt. At the same time, the temperature increase changes the liquid phase amount of the raw material that can be observed on the melting curve presented in figure 2.

For the experiment, bottom ash waste was prepared in two stages. The first stage included grinding of the powder composition in a planetary ball mill at 600 rpm for 10 min. Size fraction was 1–2 microns. Grinding provides a uniform size distribution throughout the powder composition. At the second stage, the latter was granulated in a plate granulator with addition of an aqueous polyvinyl alcohol solution sprayed on the surface of the prepared powder composition. The proportion between the solution and the powder weights was 3:7 [8]. Table 2 summarizes the parameters of obtained agglomerated powders.

Table 2. Parameters of agglomerated powders.

Raw material	Size, μm	Density, g/cm ³	Porosity, %
Ekibastuz bottom ash	100-120	0.75-0.80	40-50

Figure 3 presents images of the scanning electron microscope (SEM) investigations of ash-based porous particles which represent a set of heterogeneous oval-shaped particles. The surface of these particles is developed enough to create a branched microporous system.

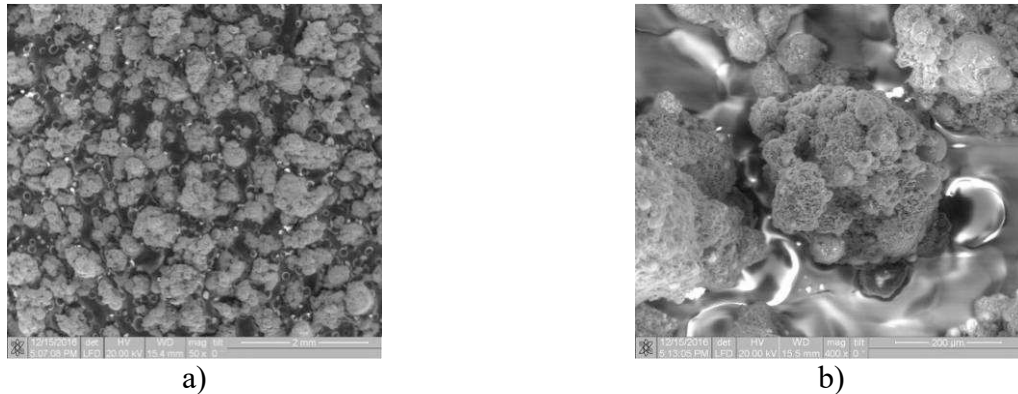


Fig. 3. SEM images of ash-based porous particles: a – 50× magnification; b – 400× magnification.

Numerical Analysis

To minimize the experimental data processing, a numerical analysis is carried out for the dynamic motion and melting of particles in the HTP flow. Specifications for the HTP flow include the heated gas flow with 10 mm radius and 70 mm length; 3200 K melt temperature; 1 l/s volume flow of plasma-supporting gas. The particles parameters include 50÷150 μm in diameter; 0–60 % porosity; 2.65 g/cm^3 density of original material (SiO_2); 743 $\text{J}/\text{kg}/\text{K}$ specific heat capacity. The porosity P of the original particle is obtained from $P = V_g/V_m$, where V_g is the pore volume in a particle; V_m is the volume of treated material; $V_g \pi D_p^3/6$ is the volume of the original particle; D_p is the particle diameter.

As a result of the numerical analysis, the dynamic motion and melting of particles in the HTP flow is obtained as shown in Fig. 4.

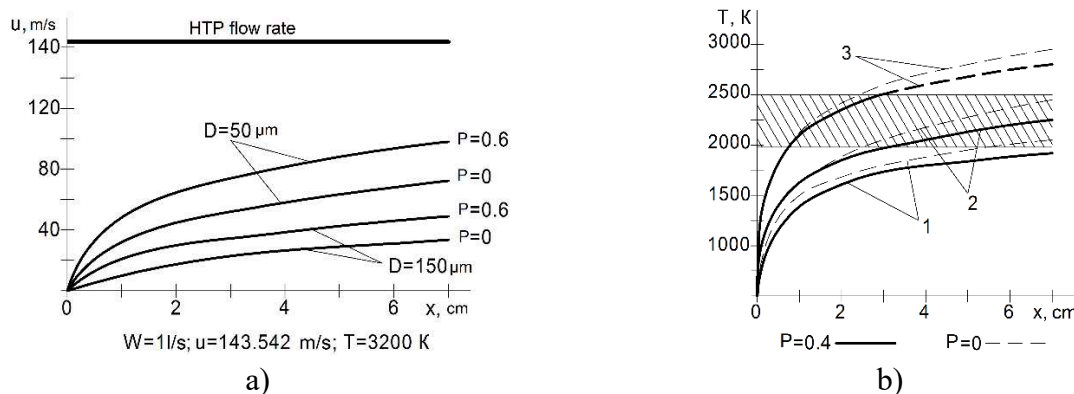


Fig. 4. Dependence diagrams for SiO_2 particles in the HTP flow:

a – dynamic motion ($\varnothing$ 50÷150 μm); b – melting; 1 – $\varnothing$ 150 μm ; 2 – $\varnothing$ 100 μm ; 3 – $\varnothing$ 50 μm .

As can be seen from Figure 4, the porosity of original particles has a significant effect on their dynamic motion and melting. It is clear that depending on the porosity of the original particle, it can be involved in different aggregative states.

Experimental

The HTP treatment of aluminosilicate porous particles was performed on a test bench which included the plasma generator (cathode), hollow graphite electrode and a chamber for collection of treated material [9, 10]. The material was conveyed from the batcher to the active zone of high-

temperature plasma flow or more precisely, plasma nozzle exit. Agglomerated particles were subjected to the intensive heating followed by their transformation to a molten state. Due to the surface tension, a heated particle took a spherical shape. Then, spherical particles travelled to the collection chamber where they instantly cooled. To avoid the aggregation of particles and facilitate their cooling, the collection chamber was filled with water.

Results and Discussions

Energy dispersive X-ray spectroscopy on the scanning electron microscope FEI Quanta 200 3D (Netherland) was carried out to investigate the morphology, shape and size of obtained spherical particles. Figure 5a presents a SEM image of aluminisilicate microspheres with 0.8 g/cm^2 density produced by the HTP treatment, while Fig. 5b contains a plot of the energy spectrum of the microsphere elemental composition.

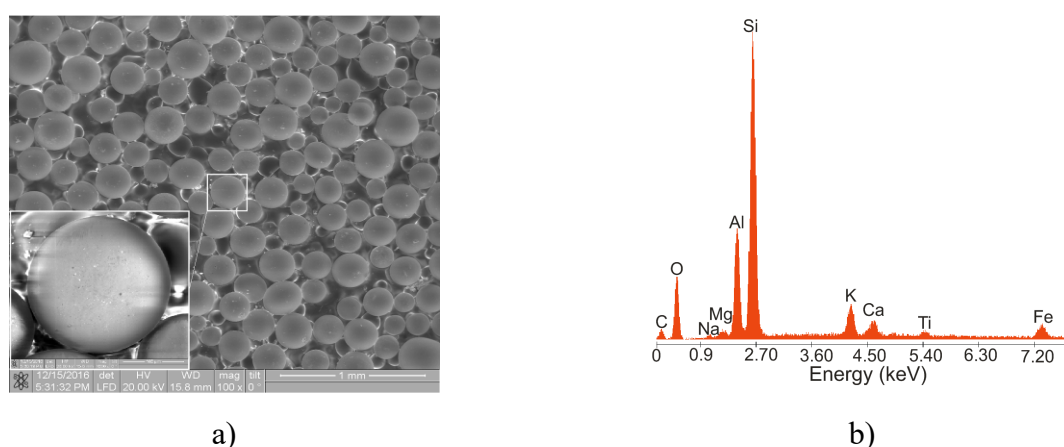


Fig. 5. Aluminisilicate microspheres: a – SEM image after the HTP treatment; b – distribution of elemental composition, wt.%: Al = 11.34; Ca = 2.63; Fe = 6.23; Mg = 0.67; Na = 0.34; Si = 34.39; Ti = 1.32; K = 5.29; O = 23.67; C = 14.10.

The investigation of the morphology of microspheres obtained by the HTP treatment shows that all of them are spherical (sphericity factor is 0.9 according to the Krumbein-Schloss diagram). The microspheres have no surface defects, however, there are aluminisilicate particles melted to their surface, which have a micron size and an acute shape.

The next stage of the experiment is the X-ray fluorescence analysis (XRF) of obtained microspheres. The elemental composition of aluminisilicate microspheres is plotted in Fig. 6. The qualitative and quantitative elemental analysis of obtained microspheres is performed on the Bruker AXS S4 Pioneer X-ray Fluorescence Spectrometer.

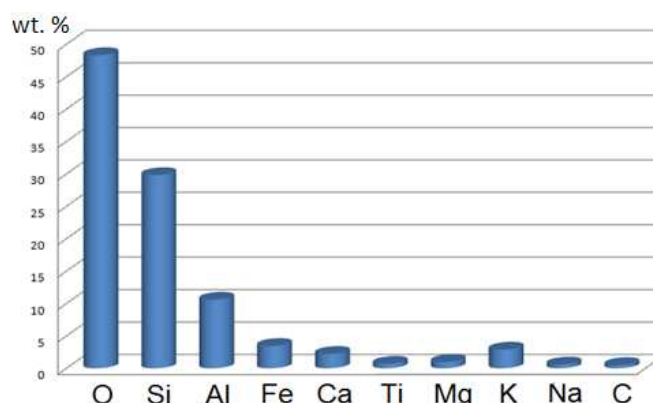


Fig. 6. The distribution of elemental composition of obtained microspheres.

In the elemental composition of obtained microspheres as shown in Fig. 3, silicon (Si) and aluminum (Al) prevail, i.e. their content is 29.82% and 10.58%, respectively, neglecting 48% of the

oxygen content. Apart from these elements, the amount of iron (Fe) is rather high (3.44%). The percentage of elements such as calcium (Ca), titanium (Ti), magnesium (Mg), kalium (K), natrium (Na) and carbon (C) is insignificant. The high content of alumina and silica indicates to a high resistance of microspheres to thermal, chemical, and mechanical effects. The results of the XRF analysis match the distribution of elemental composition on the microsphere surface shown in Fig. 5b.

Fourier transform infrared (FTIR) spectroscopy of microspheres was conducted for the investigation of intra- and intermolecular interactions and the formation of chemical bonds. FTIR spectroscopy results are presented in Figure 7 for the bottom-ash waste and microspheres produced therefrom after the HTP treatment. As mentioned earlier, bottom-ash waste is an aluminosilicate composite. So, the valence band in the absorption spectrum in 1093.82 cm^{-1} maximum is characteristic to the Si–O–Si skeleton structure. Weak maxima of 797.9 and 603.26 cm^{-1} correspond to a vibration of six-membered aluminosilicate rings in $[\text{SiO}_4]$ tetrahedrons. The absorption band at 2323.59 cm^{-1} maximum corresponds to stretch vibrations of –OH bond.

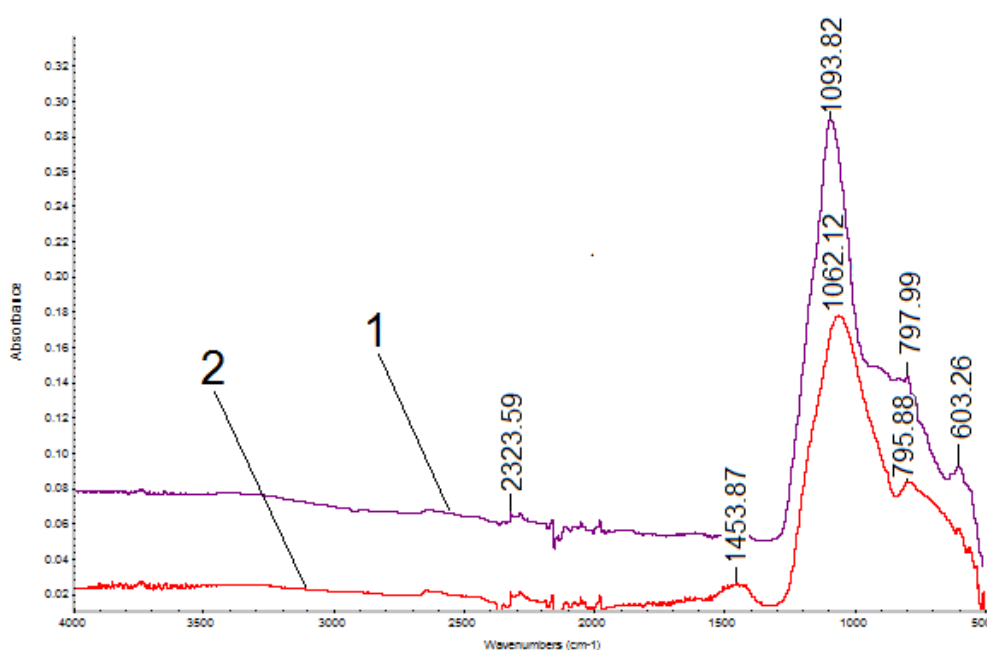


Fig. 7. FTIR spectra: 1 – bottom ash waste;
2 – aluminosilicate microspheres produced by HTP treatment.

According to FTIR spectroscopy of the original raw material and microspheres obtained by the HTP treatment shown in Fig. 7, the main absorption band shows a shift to the short-wave region resulting in 1062.12 cm^{-1} maximum. This is connected with the increase in the average distance of (Si, Al)–O and ordering of the skeleton structure due to its amorphization. Aluminum oxide is the important component which makes a distinguished contribution to the formation of the microsphere structure. Aluminum oxide has either tetrahedral or octahedral coordinates. In our case, the plasma treatment results in the reconfiguration of tetrahedral coordinates of Al cations and the increase in their content (1453.87 cm^{-1} maximum). The position of this maximum depends on the Si/Al ratio.

Conclusions

From the research that has been undertaken, it is possible to conclude that the bottom-ash waste is a promising raw material for the microsphere synthesis with the use of the plasma energy. Numerical calculations showed that the porosity of the original particle has a significant effect on their dynamic melting, and the original particle can be involved in different aggregative states while passing through 3200 K plasma flow. As a result of the experiments, microspheres with the pour density of 0.8 g/cm^3 were produced. The microspheres had no surface defects, however, they had raw material particles melted to their surface which had a micron size and an acute shape.

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Comparative Analysis of Hygiene and Sanitary Characteristics of Consumable Electrode Gas-Shielded Welding

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Keywords: welding aerosol, pulsewise supply, gas mixture, chemical composition.

Abstract. Experimental research has been carried out to determine quantitative content of solid and gas component of welding aerosol in mechanized gas-shielded welding with constant and pulsewise electrode wire supply. As a result, it has been established that the pulsewise electrode wire supply reduces the release of harmful components of welding aerosol (welding dust CO and SiO₂) by 20-25% thanks to lower energy characteristics (the transfer frequency of electrode metal droplets and the time of their exposure to critical temperatures at a short-circuit).

1. Introduction

Mechanized gas-shielded welding is a constituent part of mechanical engineering. This welding process is high-efficient, easy mechanizable and ensures sufficient mechanical properties of a weld. However, some significant drawbacks are available, one of them is hard hygiene and sanitary conditions of a welder's work, smoke content in working site, in particular [2-4].

Gas and aerosol mixture of chemical substances is generated in working site air due to cooling and condensing metal and slag vapors, as well as to their dissemination and escaping from welding zone.

A welding aerosol is a condensation aerosol according to the mechanism of its generation and is considered as a dispersion with the dispersed phase – tiny solid particles (solid component of a welding aerosol) and the disperse medium (gas component of a welding aerosol) [2].

We consider each component of a welding aerosol in more details.

Gas component of a welding aerosol is a mixture of gases resulting from thermal dissociation of gas and slag forming components in welding materials (CO, HF, SiF₄, TiF₄) or photochemical effect of welding arc ultraviolet emission on molecules of shielding gases and air surrounding the arc [2].

Solid component of a welding aerosol is synthesized due to evaporating molten electrode metal. The vapors emitted in the lower part of arc column and driven by gas flows are forced out into working site air, where they are oxidized and turned into solid particles. The role of basic metal evaporation is insignificant when generating solid component of a welding aerosol [2, 5]. It was experimentally revealed [6] that solid component of a welding aerosol depends numerically on the period of molten electrode metal drop formation, its sizes and welding arc extension. More solid component of a welding aerosol is synthesized in coarse-droplet transfer of electrode metal as against its fine-droplet transfer [6].

When transferring electrode metal with short circuits solid component of a welding aerosol is not formed intensively. Metal evaporates slightly on the edge of electrode wire and from a weld pool [1, 2]. The generating vapors are rich in manganese and other volatile substances. This process of vapor generation includes two fractions. One of them consists of low-boiling components, the other comprises low-evaporating ones [2].

Ventilation is a principal method of eliminating harmful effect of a welding aerosol on human body. On the contrary, a technological procedure necessitating appropriate welding conditions and improvement of welding technologies and materials is far more important [7, 8].

Pulsewise electrode wire supply furthers traditional welding and surfacing technological processes. It is noteworthy that the advantages of pulse arc welding as against stationary processes include [9]:

– melting, transfer and crystallization of metal can be controlled with no regard to the position of a weld pool at far lower mean values of basic technological parameters;

– crystallization of a weld pool is 2-3 times faster as a heat source effects on a weld pool in non-stationary manner and molten metal temperature decreases, as the consequence;

– deformation processes are weaker in welding constructions and on deposited surfaces;

– quality improvement of welding constructions and coating deposited when surfacing (better weld formation independently on the spatial position, increasing homogeneity of chemical composition throughout the coating, structure refinement in weld and in the heat-affected zone).

Thus, the purpose of the work is to determine quantities of solid and gas components of welding aerosol released in different methods of mechanized welding.

2. Methods and Techniques of Experimental Research

We addressed to the following welding procedures:

- mechanized gas-shielded welding with constant electrode wire supply;
- mechanized gas-shielded welding with pulsewise electrode wire supply;

For our research we suggest using (Ar+CO₂) gas mixture as a shielding gas in both welding procedures (Table 1).

Table 1. Shielding gas composition.

Welding procedure	Shielding gas
Constant electrode wire supply	82%Ar+18%CO ₂
Pulsewise electrode wire supply	70%Ar+30%CO ₂

Ar+CO₂ is used mainly due to wide industrial application of this gas shielding method, Ar+CO₂ mixture has a positive effect on technological characteristics of a welding arc (improving stability of its burning) a weld is formed better, spattering of electrode metal is reduced and fatigue life of butt welds, T-joints, and corner joints is improved 1.8 to 3.92 times [10-12].

The proportion of shielding gas components 82%Ar+18%CO₂ for welding with constant electrode wire supply is stipulated by common use of this gas content by producers of welded metal structures [12]. The choice of 70%Ar+30%CO₂ gas content (Table 1) for welding with pulsewise electrode wire supply is motivated by low spattering and the least loss of electrode metal [9].

The experimental facilities was made of automotive welding machine MECOME, equipped with a mechanism of pulsewise electrode supply [13], source of power LorchS8SpeedPulse, blending machine WittBM-2M and system of data recording Q-Data.

For the purpose of research we used 6 mm thick plates of steel (ISO) E235-B. Wire OK Autrod 12.51 (diameter 1.2mm) was applied for welding. Welding modes: $I_w=240-260A$, $U_w=24-26V$; $V_w=0.003\text{ m/s}$, $Q_g=12-15\text{ l/min}$.

Samples of air to define its dangerous level were made in the breathing zone of a welder. We used photometric method based on colorimetric reactions between metals and organic reagents. This method furthers determination of a solid component in gas-aerosol mixture of chemical substances. Gas chromatography was applied to assess gas component of a welding aerosol; this method relies on reaction gas chromatography and includes separation of carbon oxide and co-occurring components on the molecular sieve station, conversion of carbon oxide into methane and its detection by plasma-ionizing detector.

The following facilities were used:

- aspirator 822 for sampling air;
- aneroid-barometer;
- aspiration psychrometer MB-4M;
- gas analyzer "ELAN CO-50".

The aspirator was running 20 minutes each sampling, processing 0.2 m^3 of air over this period. The amount of dust generated when welding was determined as weight difference of filters used for air processing [2, 14]:

$$P=(M_1-M_2)/V \quad (1)$$

where M_1 —filter weight after welding and a check time, g;

M_2 —input filter weight, g;

V —volume of air processed by filter, m^3 .

3. Results of Research and their Discussion

In the process of experiments it was found out that constant electrode supply is distinguished by highest amount of welding aerosol components (welding dust, CO and SiO_2), both when welding in carbon dioxide and in the mixture of gases Ar+ CO_2 (Table 2).

Table 2. Quantitative chemical analysis of a welding aerosol.

Element to be measured	Unit of measurement	Welding procedure		MPC
		Constant electrode supply 82%Ar+18% CO_2	Pulsewise electrode supply 70%Ar+30% CO_2	
Welding dust	mg/m^3	3.57	2.1	-
NO_2		below 0.6	below 0.6	2.0
HF		below 0.02	below 0.02	0.5
Cr^{+6}		below 0.003	below 0.003	0.03
Cr^{+3}		below 0.5	below 0.5	3.0
Mn		0.43 ± 0.09	0.36 ± 0.10	0.6
CO		11.20 ± 0.75 ; 13.05 ± 1.5 ; 18.34 ± 1.5	11.43 ± 0.75 ; 13.31 ± 1.5 ; 13.97 ± 1.5	20.0
SiO_2		8.5 ± 2.5	6.4 ± 3.3	6.0
P_2O_5		below 0.06	below 0.06	1.0
SO_2		below 5.0	below 5.0	10.0

This is caused by better parameters of a welding procedure arc (power) (Figure 1) as a consequence, intense melting and CO emission into the arc gap and vaporous components disseminating in the environment.

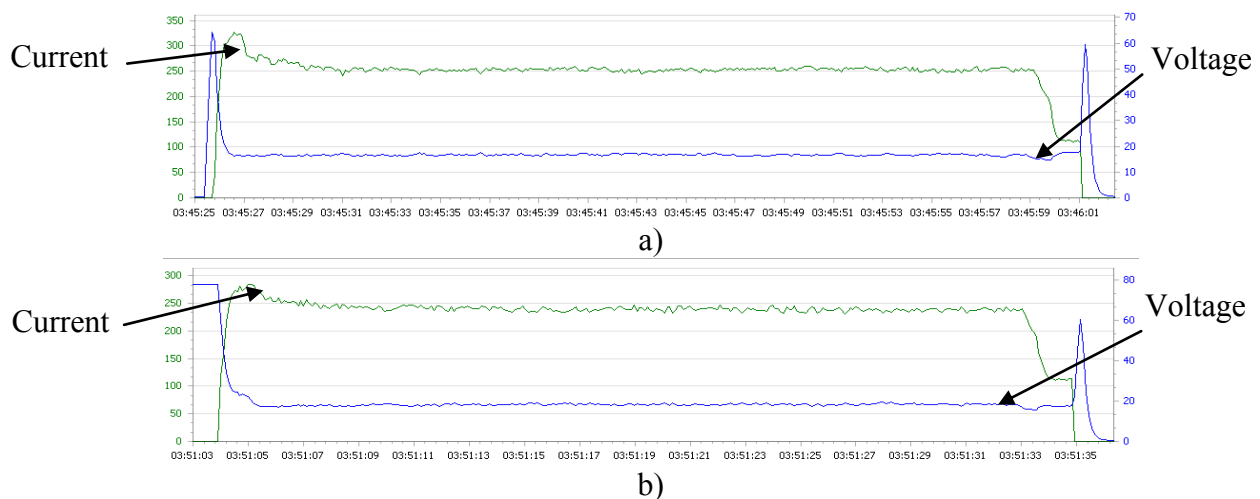


Figure 1. Welding parameters registered by the recording system Q-Data: a) constant electrode supply 82%Ar+18% CO_2 ; b) pulsewise electrode wire supply 70%Ar+30% CO_2 .

The content of welding aerosol components (NO_2 , HF, Cr^{+6} , Cr^{+3} , SiO_2 , P_2O_5 , SO_2) for the welding methods under consideration does not exceed the established maximum permissible concentrations. However, it should be noted that when welding with pulsewise electrode wire supply, the concentration of CO and SiO_2 welding dust is less by 20-25% in comparison with constant electrode wire supply welding.

The increased content of manganese in welding aerosol during constant electrode wire supply welding is caused by the elasticity of its vapors and selective evaporation from the weld pool. [7]. It depends on the time which passes from the droplet generation till the short circuit. Oscillograph patterns can give the most complete representation of the process (Figure 2).

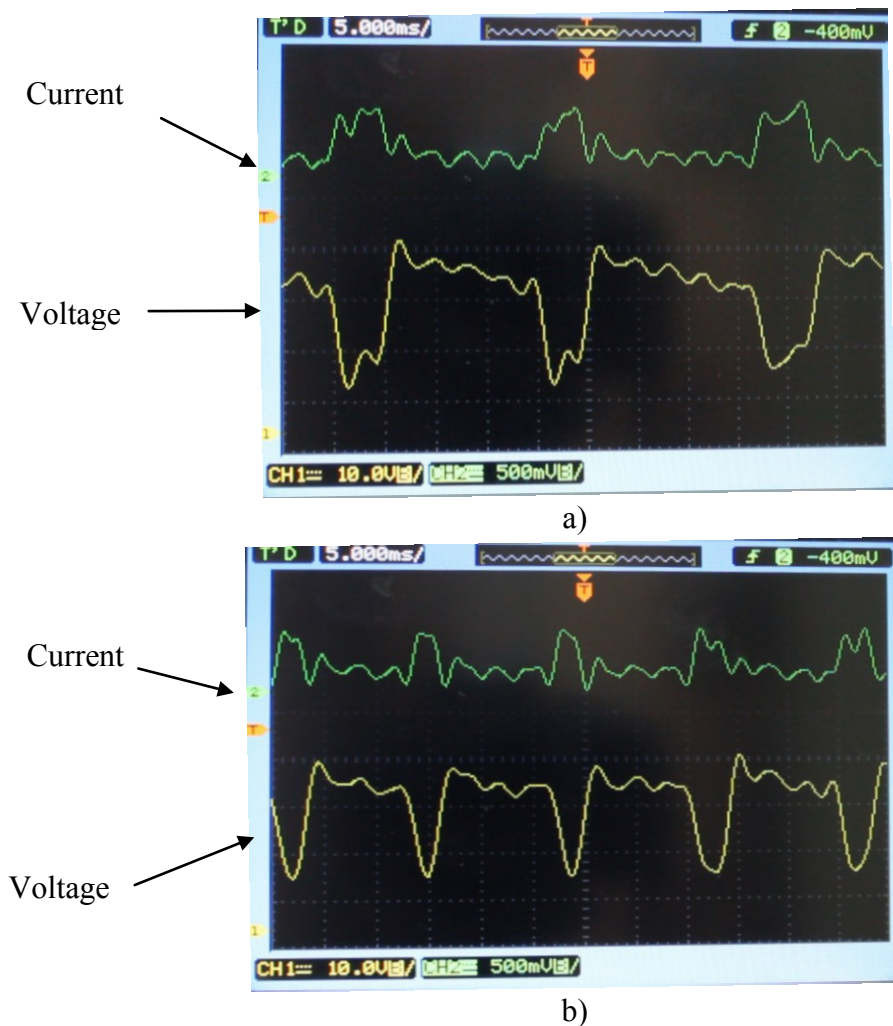


Figure 2. Oscillograph patterns of welding: a) Constant electrode wire supply 82%Ar+18%CO₂; b) Pulsewise electrode wire supply 70%Ar+30%CO₂.

As can be seen from the oscillograph patterns, for the same time interval ($t = 60$ ms), a greater number of short circuits takes place when welding with pulsewise electrode wire supply, (3 cycles for constant electrode wire supply and 6 cycles for pulsewise electrode wire supply) and, accordingly, the time of droplet transition into the weld pool as well as the droplet size decrease. Reduced electrode droplet size leads to a larger total surface of droplet contact with the surrounding medium, and hence to a more complete interaction of the liquid metal droplets with this medium (deoxidation, alloying, oxidation, gas dissolution). The time which the droplet stays in the short circuit period is also shorter (5ms for constant wire electrode supply and 2ms for pulsewise electrode wire supply). As a consequence, the droplet is subjected to critical currents during a shorter time period and correspondingly is less overheated. The combination of all these factors helps to reduce the amount of harmful welding aerosol components (welding dust, CO and SiO_2 , Mn) released into the air of the working area.

4. Conclusions

As a result of the experimental studies it was proved that the use of the pulsed electrode wire supply as a technological method for elimination of harmful influence of welding aerosol on people's health reduces the release of harmful welding aerosol components (welding dust, CO and SiO₂) by 20-25%. The high content of manganese in welding with constant electrode supply is caused by lower energy characteristics (the transfer frequency of electrode metal droplets and the time of their exposure to critical temperatures at a short-circuit).

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Influence of Gas Dynamic Processes on Chemical Composition of Hardfaced Layer when Restoring Machine Parts Manufactured from 40H Steel

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Keywords: welding, shielding gas, properties, alloyed steel.

Abstract. The authors study the influence of active shielding gas dynamics on the process of hardfacing with flux-cored wire by single-jet and two-jet gas shielding. Research results into the properties of a hardfaced layer on structural steel 40H obtained by flux-cored welding are presented.

Introduction

The problem of increasing service life of mortality machine parts is of high priority at the present time. Worn-part reclamation is an important way of increasing the economic efficiency of various machines and mechanisms [1]. For example, in Russia, expensive buckets for modern excavators by Hitachi, Komatsu and Volvo are made of a steel grade specified as 40H steel or its analogues by other producers. High-strength structural alloyed 40H steel is intended for manufacturing a wide range of products and structures subject to wear. Due to the chromium content, products from this steel have increased hardness, strength, heat resistance and corrosion resistance. Extreme operating conditions lead to rapid wear of the front wall cutting parts and the bottom of the bucket. To reduce the downtime it is necessary to study and develop an up-to-date and effective technology of repairing worn-parts which will allow restoring the specified characteristics of the part working layer as soon as possible.

Hardfacing is the most common way to restore worn parts. In industry, hardfacing can be deposited by various welding methods: manual arc welding, shielded metal arc welding (SMAW), gas metal arc welding (GMAW), vibrating electrode arc welding, electrocontact welding [2].

The most effective way to insure the necessary properties of the hardfaced layer is hardfacing by flux-cored welding [3]. Compared with solid wire, flux cored wire is considered to be a high-performance electrode material which allows solving a wide range of problems in increasing the mechanical properties of a hardfaced layer [4].

Hardfacing of excessive metal is inexpedient and leads to additional processing energy consumption. Deviation from the required properties of the weld bead is unacceptable, because it leads to a complete rejection of the product to be repaired. Therefore, in order to reduce the repair time and the cost a proper methodology for calculating and predicting the geometric parameters and properties of the weld bead is necessary. There is a number of works [5-7] which study methods for controlling the geometry and properties of hard-faced layer in the flux-cored welding hardfacing process.

A lot of technologies and devices have been developed to control welded seam forms: changing welding modes, generating mechanical impulses and electrode vibration, adding various chemical elements to the wire composition, heating the wire, introducing Al and Ni powder to the welding zone, etc. However, most methods require additional devices and equipment. A rational use of the flux-cored wire is necessary because of its high cost.

Methodology

The purpose of this paper is to study the effect of parameters and conditions of the shielding gas flow on processes in the arc discharge zone when hardfacing a bead by flux-cored welding. Hardfacing of alloyed steels has a number of special features as during the hardfacing process intensive burning out of alloying elements takes place, which significantly worsens the properties of a hardfaced layer.

To prevent burning out of alloying elements it is necessary to shorten the time during which a droplet stays on the electrode tip in an overheated state [8]. For this purpose, various pulse technologies are widely used to force a droplet detachment from the electrode tip [9]. The use of a two-jet welding nozzle results in increased action of shielding gas on the electrode metal droplet and the weld pool surface, contributes to the stability and frequency of droplet transfer which in its turn reduces overheating of the droplet and the product as a whole [10].

For reinforcing or repair hardfacing, one of the main requirements is to minimize the proportion of the base metal in the hardfaced layer in order to maximally preserve the properties of the hardfaced material of the flux-cored wire.

Power and arc pressure are the main forming factors in the hardfacing process. They basically determine the depth of penetration and the volume of the weld pool [11]. Geometric dimensions of the weld pool are influenced to a varying degree by many technological factors including a rate of shielding gas supply.

An experiment was carried out with a real research object to determine the effect of shielding gas gasdynamic properties on chemical composition of the weld metal in CO₂ consumable electrode welding. Within the experiment, a bead was welded onto a 10 mm thick plate of 40H steel. For the research we used a rutile flux cored wire of ESAB OK Tubrodur 35 GM tubing with the diameter of 1.6 mm intended for welding and hardfacing in CO₂. Due to the content of composite alloying elements, the wire makes it possible to obtain a high strength hardfaced layer allowing the product to work under intense friction and at high contact and shock loads. Therefore, it is important to choose a hardfacing method and mode insuring an increased efficiency of transmission of alloying elements from the wire to the hardfaced layer.

Beading was performed by consumable electrode welding with single-jet (conical) and with two-jet gas nozzles. Hardfacing mode: $I = 200$ A, arc voltage $U = 26$ V, welding speed $V = 3$ mm/s, welding wire feed speed $V_{pr} = 72.7$ mm/s. Shielding gas consumption Q changed from 10 to 30 l/min in increments of 10 l/min. Welding wire stick-out distance remained constant $L = 12$ mm. The equipment applied: automatic welding machine MECOME WP1500, power supply VS-300B and feeding mechanism PDGO-528M.

Results and Discussion

The content of alloying elements in the flux cored wire and in the hardfaced metal is given in Table 1.

Table 1. Chemical composition of OK Tubrodur 35 GM cored wire and hardfaced metal (mass fraction of elements in %).

Cored wire OK Tubrodur 35 GM		C	Mn	Si	Cr	Ni	Cu
		0.20	1.40	1.00	1.50	-	-
Single jet gas shielding							
Q, l/min	10	0.24	1.19	0.60	1.12	0.12	0.11
	20	0.24	1.13	0.58	1.10	0.11	0.09
	30	0.24	1.10	0.57	1.00	0.12	0.16
Two-jet gas shielding							
Q, l/min	10	0.25	1.20	0.64	1.15	0.11	0.10
	20	0.24	1.21	0.67	1.13	0.11	0.09
	30	0.25	1.15	0.67	1.18	0.13	0.11

An analysis of the experiment results has shown that a transition of alloying elements from the flux-cored wire to the weld metal is more effective when hardfacing by two-jet gas shielded welding. An increased concentration of manganese in the weld metal increases the strength, toughness, resistance to brittle fracture without reducing ductility. Silicon also increases the strength of the hardfaced layer, and in combination with manganese it provides a higher steel hardenability. One of the most important tasks is preservation of chromium in the hardfaced layer. Chromium having a high propensity to burn out significantly affects the wear resistance of a product. The higher its content in the hardfaced layer, the longer the restored part will work.

All these properties of alloying elements are extremely important for the restoration of excavator buckets operating under extreme conditions. When increasing the content of alloying elements (Figure 1) in the hardfaced metal, wear resistance and technological strength grow and hot cracks probability becomes lower [11].

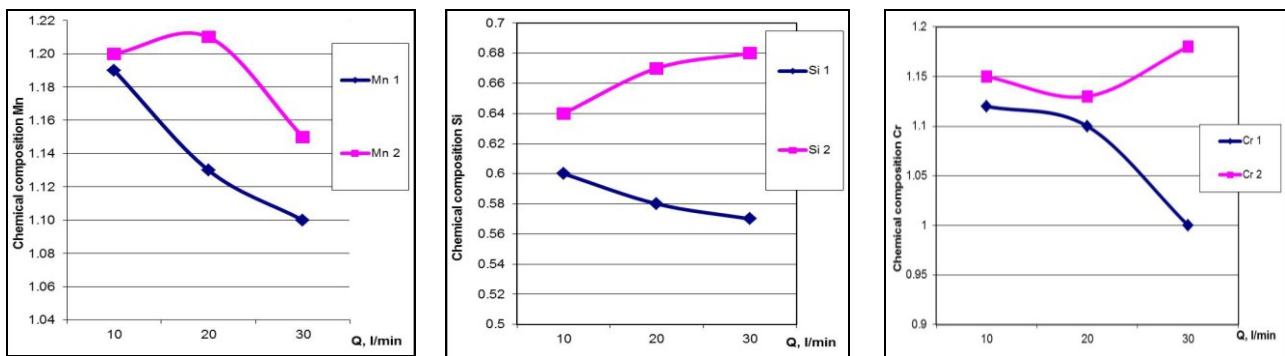


Figure 1. Changes in the content of alloying elements in the hardfaced layer depending on the shielding gas consumption: 1 - single-jet gas shielding 2 - two-jet gas shielding.

By increasing the gas flow rate, we also increase pressure of the gas jet on the molten metal in the weld pool [12]. It increases the width of the pool and decreases the penetration depth (Fig. 2) so minimizing mixing with the base metal which positively affects the properties of the hardfaced metal.

Gas flow rates at the outlet of a single-jet (conical) and a two-jet welding nozzle were measured by means of an anemometer (Table 2).

Table 2. Gas flow rate at nozzle outlet, m/s.

Gas consumption Q, l/min	Gas shielding	
	Single-jet	Two-jet
10	0.94	1.96
20	1.83	3.95
30	2.7	5.82

An analysis of changes in the penetration depth showed that the gas flow rate as a controlled parameter has a significant effect on the formation of the weld bead (Fig. 2) especially when welding with two-jet gas shielding.



Figure 2. Experimental samples with a hardfaced layer at a gas flow rate $Q = 20$ l/min: a) single-jet gas shielding, b) two-jet gas shielding.

The effect of gas dynamics on melting and electrode metal transfer during shielded metal arc welding was evaluated with the help of a high-speed video camera "Video Sprint" (Fig. 3).

Or suffer loss of information when converted to black and white, and this should be taken into account when preparing them.

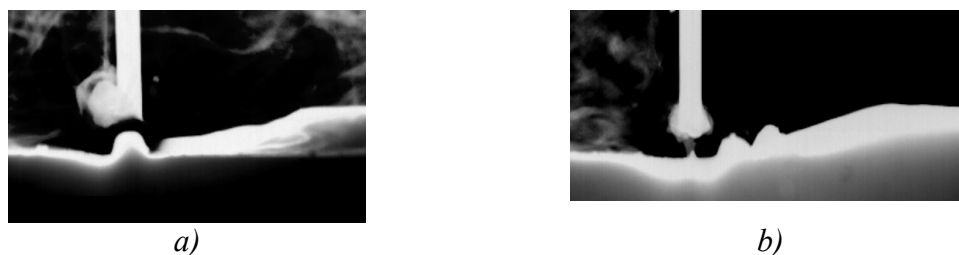


Figure 3. High-speed video images: a) single-jet gas shielding b) two-jet gas shielding.

An analysis of the high speed video images has shown that there is an increase in the gas-dynamic effect on an electrode drop in the two-jet gas shielding. The use of a two-jet nozzle makes it possible to obtain a more stable hardfacing process almost without short circuits, to reduce the size of electrode metal drops (1.5 times on average under the experiment conditions), and to reduce spattering.

When hardfacing with a flux cored wire, it is very important to achieve a simultaneous melting of the metal shell and the powder core [13, 14]. The analysis of the video images showed that when two-jet gas shielding is used melting rates of the core and the shell of the wire are equalized, frequency and stability of the drop transfer into the weld pool is increased (1.3 times on average under the experiment conditions). These factors can limit an exposure time of alloying elements of the wire to high temperature. Thus, it improves operational properties of the hardfaced layer by reducing burning out of alloying elements.

Conclusion

Two-jet gas shielding allows:

- increasing efficiency of transition of alloying elements from the flux-cored wire to the weld metal thereby improving the quality of a hardfaced layer without extra costs;
- minimizing a percentage of the base metal in a hardfaced layer;
- obtaining a weld bead of a flatter shape with minimal penetration into the base metal.

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Scanning the Welded Seams of Titanium Alloys by Using Subminiature Eddy Current Transducers

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Abstract. The eddy current transformer(EST) of the transformer type is used to construct a sensor for investigating titanium sheets joined by a welded joint. The article provides the key technical information about the eddy current transformer used and describes the methodology of measurements that makes it possible to control defects in the welding seams of titanium alloys. It is capable of automatically changing the filtering cutoff frequency and operating frequency of the device. The designed measuring system eliminates the main drawback of eddy current transformers (the small area of the electromagnetic field), simultaneously significantly reducing the noise level due to the use of high-quality amplifiers and filters, and searches for defects in and alloys of different metals. A measurement procedure that allows one to perform high-accuracy monitoring of flaws in different alloys is described. Experiments were conducted on welded BT1-0 titanium plates. The article contains the results of these measurements. The dependence data facilitates the assessment of the quality of welding seams and helps make an educated conclusion about welding quality.

Introduction

Titanium alloy is widely used in aerospace industry and military products because of its excellent over-all properties. Laser welding is greatly fitted to weld titanium alloy sheets because of its marked features such as energy concentration, small heat input, good formation of weld, purification effects and so on. However, porosities and other defects will occur in the weld because of quick cooling velocity and bad protection [1]. Because the fatigue life will reduce when there are porosities and other defects in the weld, catastrophic accident will take place in some severe states, so nondestructive testing must be done to ensure the security of important structures. A three-dimensional nondestructive testing software environment based on CAD which can rapidly and reliably evaluate the structure integrity of aircraft parts with complex structure were established by W. Deng et al [2]. Eddy current and thermography testing were adopted to detect carbon fiber intensified composite nondestructively by X. E. Gros et al [3].

Ultrasonic testing (UT) of titanium welds has low repeatability, low sensitivity, high cost and high intrusion into the manufacturing process. 20 Conventional UT would struggle to find a 2mm diameter pore in a titanium weld in a stress joint, although radiography could [4]. On the other hand, Horn et al reported being able to reliably detect subsurface pores of diameter 0.2-0.3mm diameter in 30 mm thick UNS R56401 titanium, using high sensitivity radiography[5]. Eddy Current Testing (ECT) is being considered as a replacement for the penetrant testing of titanium based components. Past experiments demonstrated that testing of titanium components can be accomplished using significantly high frequencies in the ECT range [6].

The traditional industries of industrialized countries, the cost of quality control products are 1 - 3% of the cost of production, and in industries such as defense, nuclear, aerospace, the cost of quality control of products increased to 12 ÷ 20%[7].

Despite a significant number of modern means and methods of defectoscopy of, the majority of the presented methods do not allow to scan quickly. Most of flaw detection methods may be used only in the laboratory, using complex technical instruments. Portable diagnostic methods materials

relative little functional. Despite the possibility of finding defects, they are not designed to evaluate the degradation of the material and do not allow to conclude the possible timing of its further use and the risk breakdowns.

This is due to the inability to simultaneously scan at different depths, search and analysis ultrasmall defect and scanning in real time. In this context, an urgent task is to develop eddy-current measurement systems based on miniature eddy-current transformers designed for local measurements of electrical conductivity in inhomogeneous materials, finding defects in alloys and predict the reliability of technical systems.

L. Barbato et al. [8] scanned two aluminum plates with a model flaw in the center and tested cracks between the plates. The diameter of the measuring winding was 7 mm. The scanning was performed at 1 and 5 kHz. In this case, the penetration depth of eddy currents into the studied plates at the above mentioned frequencies was 3.82 and 1.71 mm.

The analysis of a recent investigations points to the miniaturization tendency for eddy-current transformers. transformers with a size of 5×5 mm and a 0.15 mm diameter of the wire have been designed [9]. However, they do not provide the required penetration depth and localization of the magnetic field that are necessary for local measurements in different nonuniform media. Ferrite magnetic field concentrators are often used to increase the area of the magnetic field. A similar design provides an advantage that is related to the absence of the scatter of eddy currents. In addition, a 2.5 mm penetration depth is attained. In this connection, the challenge is to design eddy-current transformers that provide a penetration depth of up to 5 mm and an area of $2500 \mu\text{m}^2$. Since the eddy-current inspection method is insensitive to non conducting paint layers, it can also be used for diagnosing parts with paint coats.

Material Choices and Design

We have developed a eddy-current transformer for local monitoring of the physical parameters in titanium-alloy plates and weld seams. In contrast to existing sensors, it permits local measurements with sections measuring a few microns, to depths of the order of 5 mm. The electrical conductivity of the alloy is directly measurable, and its distribution over the sample surface and thickness may be readily established. The eddy-current method is based on the dependence of the current magnitude and distribution on the geometric and electromagnetic parameters of the sample and on the mutual position of the sensor and the sample. The basic informative parameter β_0 is a generalized characteristic of the object and the eddycurrent sensor[10]. The dimensionless parameter is defined as follows:

$$\beta_0 = D_{egu} \sqrt{\mu_0 \omega \sigma},$$

where $D_{egu} = D_1 + 1.5h$, D_1 is the mean diameter of the exciting winding; h is the distance from the equivalent turn of the exciting winding to the sample; ω is the frequency of the electromagnetic field, Hz; μ_0 is the magnetic constant; and σ is the conductivity, S. We need to develop an adequate model of the response of a eddy-current transformers of plate type, which are sensitive to many variables and permit reproduction of the voltage hodograph at small values of β_0 . We have plotted hodographs illustrating the influence of various parameters on the induced voltage on the basis of the proposed model. The eddy-current transformer is connected to the sound board of a computer with special software that controls the voltage at the transformers exciting winding and also reads the voltage at the measuring winding. On the basis of preliminary calibration, these readings are converted to values of the electrical conductivity. The windings are wrapped around a core of 2000 NMZ ferrite (relative magnetic permeability $\mu_{\text{max}} = 500$ or else of 81NMA alloy annealed by a special method. The core consists of a tetrahedral pyramid (height 1 mm), with a square base (sides of 0.2 mm).

The eddy-current transformer (Figure 1) is a transformer with measuring (1), exciting (2), and compensation (3) windings and a magnetic circuit 4, which is located inside the cylindrical platform 5 with tracks that are cut on the external side for windings. The platform is impregnated with a

compound 6 at a temperature of 200°C to prevent the disintegration of the windings when the ferrite screen 7, which is intended for the localization of the electromagnetic field on the tested object, is put in place. From the outside the transformer is contained in a corundum washer 8, which protects the core 4 from contacting the tested object.

The measuring winding rests on the points of the pyramid, which improves the localization of the magnetic field. Such transformers permit effective localization of the magnetic field, so that defects as small as 250 μm may be detected. In addition, the magnetic field penetrates into the sample to a considerable depth when working at relatively low frequencies.

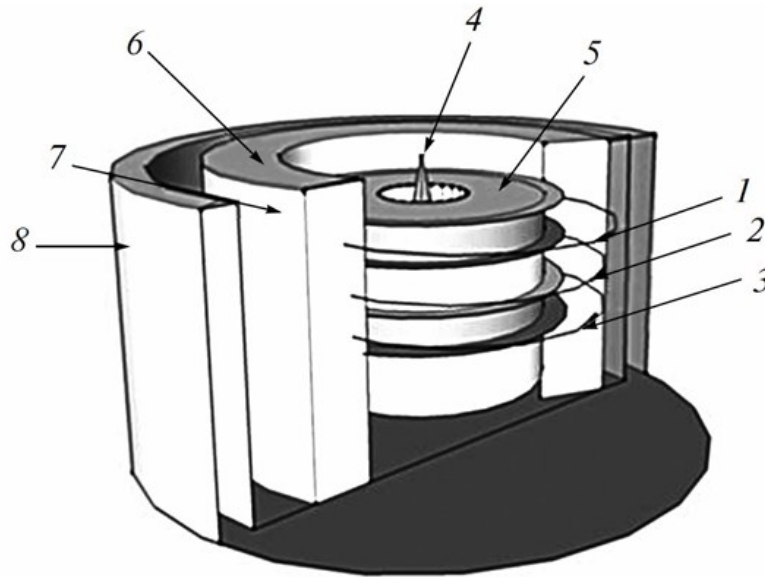


Fig. 1. Scheme of transformer

A set of measurements was performed on VT1-0 technical titanium plates that were connected by welded seams to demonstrate the appropriateness of the proposed unit for the determination of the quality of welded seams. The plate thickness was 5 mm and the width of the welded seam was 5 mm. Before the measurements were started, the transformer was calibrated. The calibration consisted of the determination of the induced voltage from the flawfree sector. Prior to the measurements, the sensor was calibrated by determining the supplied voltage from the area free from defects.

Experimental Results

In order to determine the homogeneity of the welded joints, the specimens were scanned along the surface of the welded joint. No large changes were found in the signal amplitude. The results of the experiments for the specimen No. 1 are presented in Figure 2.

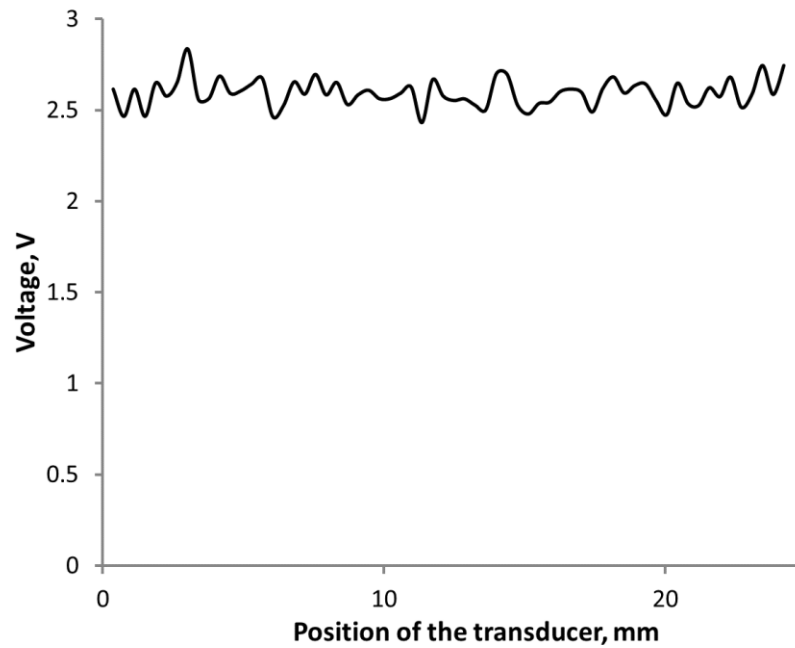


Fig. 2. The response value during the scanning of the welded seam

The experimental results indicate the relatively homogeneous structure of the welded joint but do not provide any information on the quality of the welded joint. The results of these experiments can be used to make conclusions only on the uniformity of distribution of defects along the welded joint or on the absence of defects.

In the next stage of the experiments, scanning was carried out across the welded joint. The experimental results are presented in Figures 3,4.

In the experimental specimens, the effect of a low-quality welded joint on the supplied voltage is directly indicated by the large decrease of the signal amplitude in the region of the welded joint in comparison with the region of the sheets.

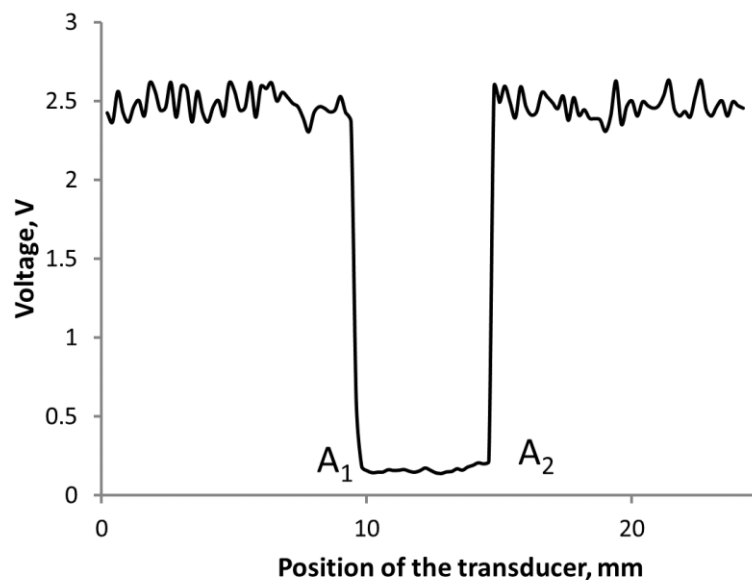


Fig. 3. The response value during the scanning of the welded seam Specimen No. 1

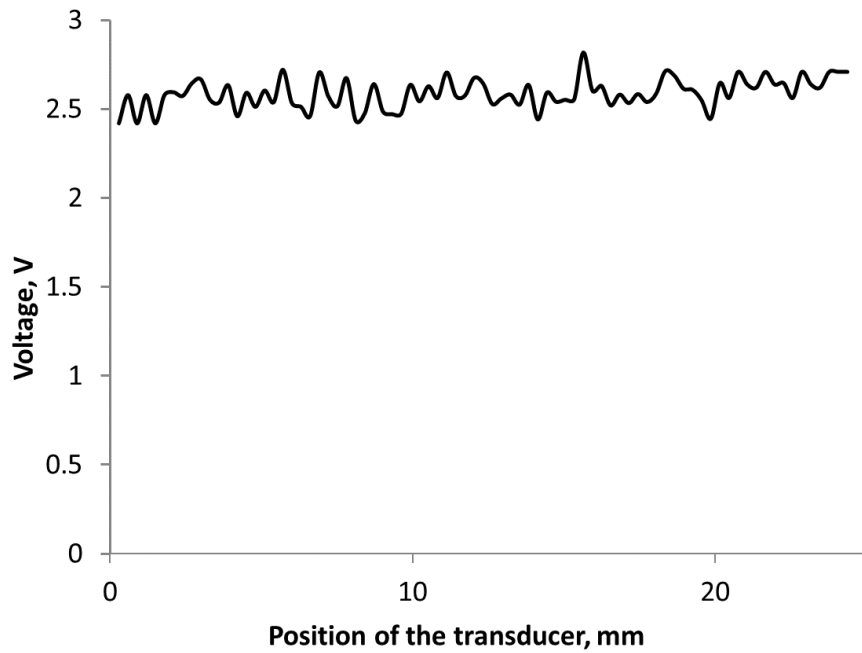


Fig. 4. The response value during the scanning of the welded seam Specimen No. 2

The experimental results were used to conclude that the quality of the welded joint in specimen No. 1 is low. The low quality of welding was also confirmed directly by sectioning the welded joint. Scanning of specimen No. 2 showed that there are no deviations of the signal amplitude within the limits of the welded joint. Examination of the sections of the welded joint in specimen No. 2 showed the high quality of the welded joint.

Additional experiments were carried out in simulation of a similar amplitude drop. In these experiments, two titanium sheets of the same thickness were placed in tight contact and subsequently the interface was scanned with a scanning frequency of 1600 Hz. The results of the experiment are presented in Figure 5.

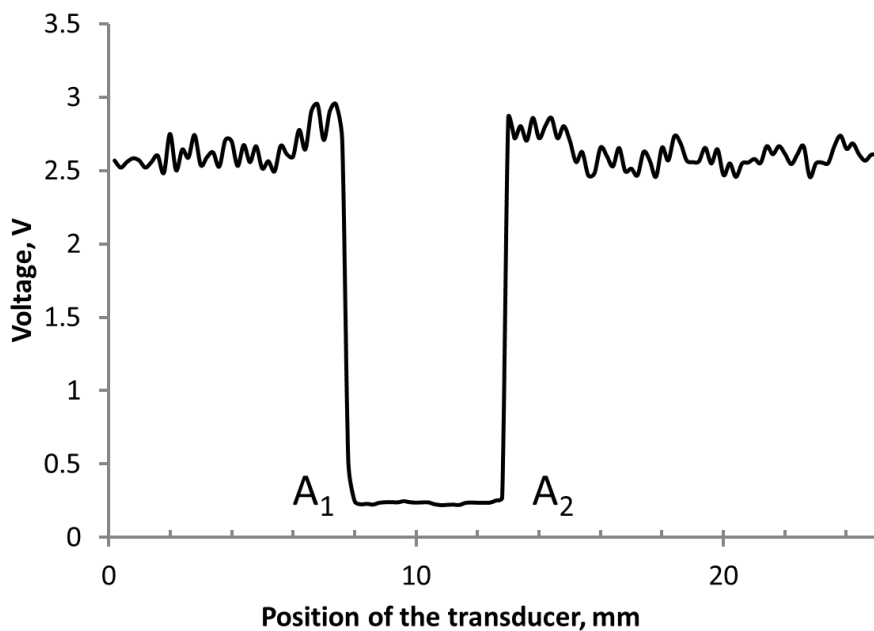


Fig. 5. The response value during the scanning in the area of two slabs junction

These experiments yielded dependence identical to that shown in Figure 3(a) in the region of the welded joint in the specimen. The signal amplitude in the region of the interface decreased by more than an order of magnitude in comparison with the signal amplitude from the sheets.

Conclusions

The experimental results show the high efficiency of the proposed method for finding defects in welded joints in titanium alloys and in evaluating the quality of welded joints. As a result of the changes in the signal amplitude of the ECT in the area of the welded joint, it is possible to determine accurately the boundaries of the welded joint and the low quality of the weld is clearly indicated by the large decrease of the signal amplitude. Similarly, the dependence obtained in scanning the second specimen with the high welding quality showed no large changes in the signal amplitude.

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