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# FLASH BUTT WELDING OF PRODUCTS OF HIGH-STRENGTH ALLOYS BASED ON ALUMINIUM

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The generalized data of technologies and equipment for flash butt welding of products of aluminium alloys and also types of products being welded using this method in rocket construction are given. The main directions of investigations for modernization of technologies of welding of high-strength aluminium alloys are considered: technology of welding using pulsed flashing as a method of heat intensification, and also effect of deformation degree at upsetting on the structure and properties of welded joints. The optimal degree of deformation of forming devices was determined allowing increase the elongation of welded product by 1.5 times at simultaneous preserving of high values of strength. In welding of thick-wall heat-hardened aluminium alloys the postweld thermomechanical treatment allows producing joints, equal in strength to base metal. Technology of welding of thick-wall T-profiles of high-strength hard-to-weld alloys was developed. The mechanical characteristics of welded joints of high-strength aluminium alloys are at the level of not less than 90 % of the strength of base metal. 10 Ref., 3 Table, 10 Figures.

# **Keywords:** flash butt welding, flashing, aluminium alloys, welded joint, thermomechanical treatment, mechanical properties, strength, ductility

The wide application of aluminium alloys for manufacture of critical structures in aircraft and rocket construction predetermined the need in development of reliable and high-efficient technologies of their joining [1]. To manufacture these structures different methods of welding are applied: arc, electron-beam, flash butt, friction, etc. [2–4]. During manufacture of critical structures of hard-to-weld alloys the permanent joints are produced using riveting. The service charac-

Table 1. Equipment for FBW of parts of alumin	ium alloys
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Machines for FBW	Maximum sections being welded, mm <sup>2</sup>	Minimum inner diameter of ring semi-products, mm	Rated power at duty cycle of 50 %, kV·A	Efficiency, welds∕h
K617	600	320	150	20
K724	600	250	100	36
K607	5000	500	350	8
K393	6000	1300	150	10
K756	15,000	900	860	6
K566	26,000	1400	930	4
K831	40,000	1700	6000	2
K754	60,000	5000	4800	4
K767	64,000	1800	4800	2
K825	90,000	1800	6000	2

2

teristics of the products depend greatly on the selected method of their joining [5].

The flash butt welding (FBW) is successfully applied at the enterprises of Ukraine and Russia for joining of different parts of high-strength aluminium alloys. The many-year experience of application of FBW evidences of high and stable quality of joints. The technological FBW process combines assembly and welding operations, does not require auxiliary consumables (electrodes, wire, fluxes, shielding gases) and easily adapted to automation and robotization. In flash welding the precision preparation of edges of parts is not required [6–8].

At the E.O. Paton Electric Welding Institute of the NAS of Ukraine (PWI) the technologies and equipment for FBW of different products of alloys based on aluminium with the cross section area of up to 90,000 mm<sup>2</sup> (Table 1) were developed. The technology and specialized equipment are implemented at the plants of rocket industry in Ukraine and Russia.

The FBW technology is applied for joining of:

• products of closed shape (frame rings) of pressed profiles with intricate and differentthickness cross section of up to 60,000 mm<sup>2</sup> area (the examples of profiles being welded are given in Figure 1);

 $\bullet$  longitudinal welds of shells of bodies of fuel tanks with section of up to 2000  $\times$  32 mm of

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alloys AMg6NPP, 1201, and 2000  $\times$  45 mm of alloy AMg6M;

• longitudinal load-carrying set (stringers, fittings) of thickness from 2 mm of aluminium alloys of different alloying systems (V95, D16, AK6) in similar and dissimilar combinations.

FBW of frame rings with cross section area of 2500 mm<sup>2</sup> using machine K393 and FBW of a shell with cross section area of  $64,000 \text{ mm}^2$  using machine K767 are shown in Figure 2.

In connection with wide application in welded structures of high-strength aluminium alloys, relating to the category of hard-to-weld, the need in modification of FBW technology arises. It was found that to obtain high values of mechanical properties of welded joints it is necessary first of all to optimize the heat input in welding and to control the temperature distribution in HAZ metal. In welding of heat-hardened alloys it is especially important to optimize the temperature and time of heating [9]. Alongside with this the sufficient influence on mechanical properties of joints is exerted by the conditions of deformation of near-contact layers of metal subjecting to intensive deformation.

As a result of many-year investigations at the PWI the methods of heat intensification were developed (program decrease of voltage, pulsed



Figure 1. Examples of aluminium alloy profiles welded by  $\ensuremath{\mathsf{FBW}}$ 

and pulsating flashing) [6]. At the present time the technologies of heating using pulsating flashing of aluminium alloys with application of modern powerful hydraulic drives and computer control systems are developed at the PWI. The principle of the method of heating using pulsating flashing consists in maintenance of welding current in the range, at which the highest efficient power is generated in the contact between two parts. The given mode is provided by control of welding speed depending on the values of welding current. The application of pulsating flashing instead of preheating by resistance heating allows producing uniform high-concentrated heating





Figure 2. FBW of frame rings in machine K393 (a) and shells in machine K767 (b)



**Figure 3.** Program of change of main parameters of process in welding of aluminium alloy parts of cross section area  $30 \times 140$  mm by pulsating (*a*) and continuous (*b*) flashing



Figure 4. Scheme of formation of welded joint with end gap  $\Delta_{end}$ 



Figure 5. Dependence of deformation degree, bending angle of fiber and elongation on end gap

across the welded section, decreasing tolerances for flashing and duration of heating.

The carried out investigations showed that the possibility of smooth control of welding voltage in heating using pulsating flashing allows presetting the mode of heating with the most optimal welding parameters (Figure 3).

To increase the plastic characteristics of welded joints the influence of deformation degree

**Table 2.** Mechanical properties of base metal and welded joints of alloy 1201 of 30 mm thickness

Material	ε, %	$\sigma_t$ , MPa	δ <sub>5</sub> , %	$K = \sigma_{\rm t \ WJ} / \sigma_{\rm t \ BM}$
Base metal	_	433	15	-
Welded joint	93	411	7	0.95
	80	427	10	0.98
	67	430	11	0.99

on mechanical properties of welds during formation of welded joint was investigated.

In welding with forming devices (Figure 4) the metal of thickness  $\delta$  is extruded into the gap  $\Delta_{g}$  during upsetting, the value of which is changed with time, the deformation degree  $\varepsilon$  is also changed with time [10]:

$$\Delta_{\rm g}(t) = \Delta_{\rm end} + \Delta_{\rm ups} - v_{\rm shr}t, \quad \varepsilon(t) = \frac{\delta - \Delta_{\rm g}(t)}{\delta} \ 100 \ \%,$$

where  $\Delta_{ups}$  is the tolerance for upsetting;  $v_{shr}$  is the speed of shrinkage.

With the increase of end gap between the forming devices the deformation degree and bending angles of fibers under the edges of knives decreases. It was found that in flash welding the decrease of deformation degree in welded joint due to increase of end gap allows increasing the elongation  $\delta_5$  from 7 to 10–11 % at simultaneous preserving of high values of strength (Figure 5). The macrostrucutres of the zone of welded joint with different end degree of deformation are given in Figure 6. The results of mechanical tests of welded joints obtained in FBW, specimens of 1201 alloy of 30 mm thickness at different conditions of extrusion are given in Table 2. The application of thermomechanical treatment after welding allows producing welded joints close to the strength of base metal (Table 3).

The development of welding technology was accompanied with modification of design of FBW equipment. The systems of quick-response hydraulic drives of machines on the base of modern complexes, computerized systems of control of parameters of flashing process were developed. They allow reproducing the welding modes (see Figure 3, a) at high accuracy, providing stable flashing at higher densities of current than at continuous flashing (see Figure 3, b).

In the process of optimizing the welding modes the criterion for quality evaluation under industrial conditions is the bend test of welded joints, having preliminary notch in the weld, up to fracture (quality express-analysis). This method allows visual detecting of defects directly in the

Table 3. Mechanical properties of base metal and welded joints without and with postweld thermomechanical treatment

Alloy (analog)	W	ithout heat treatme	ent	Postweld	Postweld thermomechanical treatment						
	$\sigma_t$ , MPa	σ <sub>0.2</sub> , MPa	δ <sub>5</sub> , %	$\sigma_t$ , MPa	σ <sub>0.2</sub> , MPa	δ <sub>5</sub> , %	R				
1201 (2219)	176/177	86/100	23/23	441/432	325/345	25/8	0.98				
V95 (7075)	217/219	103/128	21/13	487/479	419/406	12/9	0.98				
AD33 (6061)	132/131	67/73	32/27	281/301	276/297	19/15	1				
Note. In numerat	Note. In numerator the values for base metal are given, in denominator — for welded joint.										



Figure 6. Macrostructures of zone of welded joint (a-c) in welding alloy 1201 of 30 mm thickness at different final degree of deformation

weld. The example of fracture is given in Figure 7.

In the structure of welded joints the following is distinguished:

• base metal with large grains elongated along the line of rolled metal and large clusters of intermetallics along their boundaries;

• HAZ which consists of areas with different level of recrystallized and deformed metal that leads to sharp decrease of grains size and their orientation relatively to the direction of grains of base metal up to  $90^{\circ}$ ;

• weld metal with dense fine-grain structure.

The examples of microstructure of welded joints, produced using FBW method, of speci-



Figure 7. Fractures of welded joints



Figure 8. Macrostructures of welded joints of alloys 1201 (a), AD33 (b), 1570 (c) and V95 (d) made by FBW

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Figure 9. Riveted and welded joint of dissimilar alloys on aluminium base of fitting-stringer elements used in aerospace industry

mens of 30 mm thickness of alloys 1570, AD33, 1201 and V95 are given in Figure 8.

The application of welding instead of riveting is one of the efficient methods of solution of problem of decrease of weight of flying vehicles (Figure 9).

As a result of conducted investigations the FBW technology of thin-wall heat-hardened T-profiles of V95-T1 alloy was developed applied in manufacture of assemblies of longitudinal load-carrying elements of the last stage at efficient loading of rocket-carriers (fitting-stringer and stringer-stringer). This technology provides coefficient of strength of welded joints  $K \ge 0.9$ . The average values of mechanical properties of base metal (numerator) and welded joints (denominator) of profile 2.5 mm thick of V95-T1 alloy are given below:  $\sigma_t = 580/536$  MPa;  $\sigma_{0.2} = 505/426$  MPa;  $\delta_5 = 15/6$  % (K = 0.92).

Basing on the FBW technology of thin-wall profiles of high-strength aluminium alloys the new machine for welding of stringer–fitting, stringer–stringer elements (Figure 10) was developed providing welding of profiles of high-strength aluminium alloys of up to 300 mm<sup>2</sup> cross section area.

The application of this technology in production for welding of fitting-stringer joints instead of riveted ones, being used at the moment, will allow sufficiently increasing the weight of useful load delivered to the orbit by domestic rocketcarriers.

#### Conclusions

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1. The use of flash butt welding technology for joining of parts of high-strength alloys based on aluminium provides mechanical properties of welded joints at the level of not lower than 90 % of strength of base metal.



**Figure 10.** Machine for FBW of thin-wall profiles of highstrength aluminium alloys

2. The optimal degree of deformation at upsetting using forming devices was developed, which allows increasing the elongation of welded product by 1.5 times at simultaneous preserving of high values of strength.

3. In welding of thick-wall heat-hardened aluminium alloys the postweld thermomechanical treatment allows producing joints, equal in strength to the base metal.

4. The technology and equipment for FBW of thin-wall profiles of high-strength heat-hardened alloys based on aluminium were developed.

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# FORMATION OF COLD CRACKS IN WELDED JOINTS FROM HIGH-STRENGTH STEELS WITH 350-850 MPa YIELD STRENGTH

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The main problems in welding of high-strength steels are connected with their susceptibility to cold crack formation. As a rule such cracks are nucleated in HAZ of welded joints under the effect of tensile stresses. Diffusible hydrogen and presence of hardening structures in metal accelerate this process. Given paper presents a comparative analysis of effect of structure, diffusible hydrogen and residual stresses on cold crack resistance of welded joints from high-strength structural steels differ in chemical composition and level of static strength. Experiment-calculation methods of investigation were used for study of micros-tructural changes and formation of stress-strain state in rigidly fixed welded joints. Resistance of welded joints to cold crack formation was evaluated based on results of testing of technological samples and specimens on Implant method. It was determined as a result of performed investigations that probability of formation of longitudinal cold cracks in rigidly fixed welded joints from high-strength steel changes in wide ranges. However, there are specific regularities related with effect of the residual welding stresses on this processes. Increase of diffusible hydrogen content in deposited metal, steel carbon equivalent, cooling rate and stress-strain state of welded joints reduce their cold crack resistance. Results of performed investigations can be used in development of technological processes of welding of high-strength steels with yield strength from 350 to 850 MPa and carbon equivalent from 0.35 to 0.70 %. 11 Ref., 4 Tables, 6 Figures.

**Keywords:** high strength low-alloy steels, welded joints, cold cracks, residual welding stresses, diffusible hydrogen, metal structure, preheating, heat-affected zone

The main problems in welding of high-strength steels are connected with their susceptibility to formation of cold cracks [1-5]. It is well known fact that presence of hardening structures, diffusible hydrogen  $[H]_{dif}$  and tensile stresses [1, 3, 6, 7] in the place of crack potential nucleation are necessary conditions for formation and development of cold crack in welded joints.

Mostly, effect of indicated factors on cold crack resistance of welded joints is evaluated on results of indirect investigations. For this, relation between technological parameters of welding and structure of weld metal and HAZ of welded joints, conditions of saturation of deposited metal by [H]<sub>dif</sub>, capability of metal to resist continuous external tensile stress without failure is investigated. However, the question to what extent conditions of such tests correspond to processes taking place in the welded joints is not enough studied. Meanwhile, development of experimental as well as calculation methods for determination of distribution of parameters of indicated characteristics in welding of different combinations of structural steels allows more accurate selection of technological procedures directed on prevention of risks of cold crack formation in welded joints.

In this connection, aim of the present paper lied in evaluation of effect of hardening structures, [H]<sub>dif</sub> and residual stresses on cold crack resistance of welded joints from high-strength structural steels using modern experiment-calculation methods of investigation as well as results of testing of specimens on Implant method.

Mathematical modelling of processes of microstructural changes and formation of stressstrain state in rigidly fixed welded joints makes a basis of the experiment-calculation method.

Well approved algorithm [8], according to which weight fraction of ausenite  $V_{\rm a}$ , bainite  $V_{\rm b}$ and martensite  $V_{\rm m}$  in total always equals one was used for mathematical description of microstructural changes, increment of free deformations and change of yield strength of the metal depending on calculation data with respect to temperatures in arbitrary point (x, y, z) in moment of time t.

Residual content of martensite, bainite and austenite was determined on time of staying of (x, y, z) point in 800–500 °C ( $\Delta t_{8/5}$ ) temperature range in accordance with dependences, given in work [9].

Level of residual stresses in discrete points of welded joint was determined by finite element



**Figure 1.** Scheme of technological sample of  $\delta$  thickness and 2*L* width, fixed by flange welds to plate of  $\delta_p >> \delta$  thickness with transverse (1) and longitudinal (2, 3) cracks in root weld

method based on corresponding mathematical models, developed by Vladimir I. Makhnenko at the E.O. Paton Electric Welding Institute [10].

Probability of nucleation of cracks in welded joints from steel depending on content of  $[H]_{dif}$ in deposited metal (regulated by conditions of electrode baking), residual stresses (regulated by change of specimen fixing basis) and structural change of HAZ metal (regulated by rate of specimen cooling due to change of initial temperature of the plate  $T_0 = 11$ , 60 and 80 °C) was evaluated on results of testing of technological samples (Figure 1). 10 single-type specimens were tested at each specific welding conditions. Tests on Implant method were carried out in accordance with procedure described in [1]. Cylinder specimen-inserts of 6 mm diameter with notch in a form of spiral grove (step 1.25 mm, radius near the top 0.1 mm) were used.

Conditions of formation of cold cracks in manual metal arc (MMA) welding of root weld of rigidly fixed butt joints of the technological samples with fixing basis L = 50, 70 and 100 mm from steel 14KhG2SAFD of  $\delta = 18$  mm are considered taking into account mentioned general provisions. MMA welding of specimens was carried out using ANP-10 electrodes of 4 mm diameter. Parameters of welding mode are the following: direct current  $I_w = 140-150$  A; arc voltage  $U_a = 24$  V;  $v_w = 7.2-7.5$  m/h. Table 1 shows chemical composition of the base and filler materials.

Figure 2 shows dependences characterizing changes of  $V_{\rm m}$  in sections of welded joint located at x = 0.25 and 5 mm height from root surface of the weld at different initial temperatures of metal.

As follows from Figure 2, cooling rate of welded joint at  $T_0 = 11$  °C makes  $w_{6/5} \approx 25-30$  °C/s. Quantity of martensite is around 90 % in zone of potential crack formation (area of HAZ metal overheating in Figure) or in point with x = 0.25 mm, y = 2.5 mm (Figure 2, *a*) coordinates. Heating to 70, 90 and 120 °C provides 72, 65 and 50 % decrease in  $V_{\rm m}$  due to reduction of



**Figure 2.** Calculation  $V_{\rm m}$  values in cross section of root weld of studied sample at x = 0.25 and 5 mm height (*a* and *b*, respectively) with L = 50 mm and  $T_0 = 11$  (1), 70 (2), 90 (3) and 120 (4) °C

Material	С	Si	Mn	Cr	Cu	V	Ab	Р	S	$P_{\rm cm}$
Steel 14KhG2SAFD	0.13	0.57	1.42	0.44	0.39	0.08	0.08	0.019	0.015	0.27
Electrode ANP-10	0.09	0.43	1.90	_	_	0.01	-	0.020	0.020	0.20

 Table 1. Chemical composition of base and filler materials, wt.%

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 $w_{6/5}$  up to 10 °C/s. Thus, presence of specific quantity of hydrogen in metal and stresses in given point of welded joint develop all the conditions for formation of cold cracks.

Figure 4 represents the data on distribution of longitudinal  $\sigma_{zz}$  and transverse  $\sigma_{yy}$  stresses in zone of cold crack formation of studied welded joints depending on specimen fixing basis. They indicate that level of residual stresses in welded joints of the technological samples changes in wide ranges, approximately from 400 to 120 MPa in the weld metal and from 590 up to 390 MPa in HAZ with increase of *L* from 50 up to 100 mm.

Table 2 shows the results of testing of technological samples from steel 14KhG2SAFD. They indicate that longitudinal cold cracks will be formed with probability *P* around 1.0 in welding of root weld of the technological samples with L = 50 mm and at  $T_0 = 10$  °C, when [H]<sub>dif</sub> content in the deposited metal is on the level of 7 ml/100 g and above. Reduction of [H]<sub>dif</sub> up to 5.5 ml/100 g decreases *P* to 0.7, and *P* = 0 at [H]<sub>dif</sub> = 40 ml/100 g. Probability of crack formation in studied samples also reduces with increase of welded joints *L*, namely at *L* = 70 mm up to *P* = 0.5, and to *P* = 0 at *L* = 100 mm.

As can be seen from Figure 2, content of martensite in metal near the fusion zone in area of HAZ metal overheating reduces from  $V_{\rm m} = 90$  % at  $T_0 = 11$  °C to  $V_{\rm m} = 50$  % at  $T_0 = 120$  °C with  $T_0$  increase. However, it is difficult to evaluate a real effect of  $T_0$  on probability of cold crack appearance in the root layer of welded joints from 14KhG2SAFD steel based on these data. Additionally, data on principal  $\sigma_{zz}$  and transverse stresses  $\sigma_{yy}$  should be analyzed for this.

Calculations show that maximum values of  $\sigma_{zz}$  and  $\sigma_{yy}$  reduce, approximately, from 590 MPa



Area of overheating of HAZ metal

**Figure 3.** Cross microsection of root weld in studied sample at L = 50 mm,  $T_0 = 11$  °C and [H]<sub>dif</sub> > 7 ml/100 g

at  $T_0 = 11$  °C up to 500 MPa at  $T_0 = 120$  °C in HAZ area with  $T_0$  rise at L = 50 mm. It is also characteristic that this effect, that is partially caused by reduction of maximum stress values and partly by decrease of quantity of points (volume) with high  $\sigma_{zz}$  and  $\sigma_{yy}$  values, determines lowering of probability of cold crack formation with  $T_0$  rise.

Generalized data concerning effect of [H]<sub>dif</sub> content in the deposited metal and level of residual stresses  $\sigma_{yy}$  on cold crack resistance of butt technological joints of samples from 14KhG2SAFD steel at  $V_{\rm m} \approx 90$  % are given in Figure 5 (curve 1). This Figure also shows for comparison the results of specimen testing on Implant method (curve 2), welding of which was carried out using the same modes as for root weld of the technological sample. Due to this cooling rate of specimen-insert was correlated with cooling rate of HAZ metal of the welded joint  $(w_{6/5} = 25 \text{ °C/s})$  and structure consisting of martensite and bainite (90 and 10 %, respectively) was formed in it under the effect of welding thermal cycle.



**Figure 4.** Calculation values of  $\sigma_{zz}$  (*a*) and  $\sigma_{yy}$  (*b*) in cross section of root weld at  $T_0 = 11$  °C, x = 0.25 mm and L = 50 (1), 70 (2) and 100 (3) mm



**Figure 5.** Cold crack resistance of HAZ metal of welded joints of 14KhG2SAFD steel based on results of technological samples (*t*) and specimens on Implant method (*2*)

Thus, investigations, performed as applied to MMA welding of root weld of rigidly fixed welded joints from 14KhG2SAFD steel 18 mm thick, showed that probability of formation of longitudinal cold crack in them changes in wide ranges at selected welding mode and varying  $T_0$  from 11 up to 120 °C, content of [H]<sub>dif</sub> from 4.0 to 8.6 ml/100 g and L from 50 to 100 mm. How-

ever, there are dependences related with effect of residual welding stresses on this process. Similar dependences were detected in testing of specimens on Implant method.

Considering that the results of testing of rigid technological samples and specimens, investigated on Implant method, showed good comparability, further investigations, directed on comparison of susceptibility to cold crack formation of series of domestic high-strength steels with different chemical compositions and indices of static strength, were performed using the same method.

Chemical compositions of studied steels are given in Table 3.

Relative value  $\sigma_{cr} / \sigma_{0.2}$  was accepted as a criterion providing more specific comparison of susceptibility to cold crack formation of specified steels taking into account that strength indices in the studied steels are significantly differ between each other. Index  $\sigma_{cr}$  was determined based

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 $\begin{array}{l} \textbf{Table 2.} \\ \textbf{Results of investigations of technological samples from steel 14 KhG2 SAFD 18 mm thick made by MMA welding using ANP-10 electrodes \end{array}$ 

			L, mm											
Content of $[H]_{dif}$	50	70	100	50	70	100	50	70	100	50	70	100		
metal , ml/100 g		Quantity of specimens with cracks the length of which correspond to specimen length, pcs			Quar fractur	Quantity of partially fractured specimens, pcs			Quantity of specimens without cracks, pcs			Probability of formation of cracks <i>P</i>		
4.0	11	0	0	0	0	0	0	10	10	10	0	0	0	
5.5		3	0	0	$4^{*}$	0	0	3	10	10	0.7	0	0	
7.0		10	0	0	0	0	0	0	10	10	1	0	0	
8.6		10	4	0	0	2**	0	0	4	10	1	0.6	0	
8.6	60	5	0	0	1***	0	0	4	10	10	0.6	0	0	
8.6	80	0	0	0	0	0	0	10	10	10	0	0	0	
*Length of crack i	in relation	n to specir	nen lengt	h made 80	), 55, 40 a	und 35 %.	**The sam	ne, but 30	and 70 %	5. *** The s	ame, but	60 %.		

Table 3. Chemical composition (wt.%) and carbon equivalent  $C_{eq} \mbox{ of studied steels}$ 

Steel	С	Si	Mn	Cr	Ni	Mo	Cu	V	Nb	Al	S	Р	C <sub>eq</sub> , %
10KhSND	0.09	0.98	0.70	0.77	0.80	_	0.37	_	_	_	0.018	0.020	0.46
09G2SYuch	0.01	0.36	1.90	_	_	_	0.39	_	_	0.06	0.010	0.015	0.44
06G2B	0.08	0.27	1.50	_	_	0.19	0.23	_	0.05	0.04	0.006	0.011	0.37
14G2GMR	0.15	0.28	1.10	1.30	-	0.43	0.20	-	0.02	0.05	0.023	0.024	0.63
12GN2MFAYu	0.15	0.41	1.14	0.38	1.56	0.22	0.19	0.07	-	0.06	0.032	0.014	0.50
14KhGN2MDAFB	0.14	0.25	1.30	1.15	1.94	0.24	0.42	0.14	0.04	0.05	0.008	0.014	0.70
12GN3MFAYuDR	0.13	0.23	1.36	_	3.08	0.33	0.40	0.05	_	0.02	0.004	0.020	0.48

					<i>w</i> <sub>6∕5</sub> , °C∕s					
Steel		5			10		25			
	σ <sub>0.2</sub> , MPa	$\sigma_t$ , MPa	δ <sub>5</sub> , %	σ <sub>0.2</sub> , MPa	$\sigma_t$ , MPa	δ <sub>5</sub> , %	σ <sub>0.2</sub> , MPa	σ <sub>t</sub> , MPa	δ <sub>5</sub> , %	
06G2B (465, 530, 30)	440	620	28	445	640	28	490	645	26	
9G2SYuch (460, 590, 31)	420	630	30	430	640	28	470	650	26	
10KhSND (470, 610, 31)	480	720	26	510	740	24	560	790	22	
12GN2MFAYu (625, 720, 20)	780	890	18	810	950	16	890	1050	16	
14KhG2GMR (680, 780, 18)	900	1040	17	1000	1180	16	1050	1180	14	
14KhGN2MDAFB (860, 920, 17)	1100	1150	12	1050	1300	12	1100	1350	12	
12GN3MFAYuDR (821, 887, 19)	800	960	16	920	1050	14	1050	1210	13	
Note. Values of $\sigma_{0,2}$ , $\sigma_t$ (MPa) and $\delta_t$	<sub>5</sub> (%) respect	ively for stu	died stee	ls in initial c	ondition.					

 Table 4. Mechanical properties of metal of simulated HAZ of studied steels

on results of Implant method and  $\sigma_{0.2}$  values (conditional yield strength of HAZ metal) was obtained during static tension tests of specimens of II type on GOST 6996, manufactured from metal billets and treated using specific thermal welding cycle on MSR-75 machine [6]. Table 4 shows the results of these investigations.

Given results verify that steels of different chemical composition have different reaction on effect of thermal cycle. Low-alloy steels of 06G2B, 09G2SYuch and 10KhSND grade have the lowest susceptibility to hardening. Insignificant weakening can be observed in HAZ metal of such steels at low cooling rates ( $w_{6/5} \leq 10$  °C/s). As for high-strength alloy steels, they have typical increase of strength indices in HAZ metal even at low cooling rates. In particular, this refers to steels containing chromium (10KhSND. 14Kh2GMR and 14KhGN2MDAFB) which, as everybody knows, increases steel susceptibility to hardening.

Figure 6 shows the dependencies characterizing susceptibility of high-strength steels of different strength grades to cold crack formation.

It is determined that welded joints from 06G2B steel have the highest delayed fracture resistance independent on cooling conditions. Even if hydrogen concentration reaches the limit, HAZ metal of such steels shows no susceptibility to cold crack formation. Such a high resistance of specified steel to cold crack formation can be explained by very low values of  $C_{eq}$ , around 0.37 %.

Low-alloy steels 09G2SYuch and 10KhSND ( $C_{eq} \approx 0.44$  and 0.46 %, respectively) also have high cold crack resistance. However, in contrast to 06G2B steel, they are more sensitive to hy-

drogen embrittlement and require either more rigid limitations on  $[H]_{dif}$  saturation of welds or  $w_{6/5}$  delay due to preheating application.

HAZ metal of welded joints from highstrength steels of 12GN2MFAYu and 12GN3MFAYuDR type has cold crack resistance comparable with low-alloy steels 09G2SYuch and 10KhSND at limited content of [H]<sub>dif</sub> and delayed cooling. But, since their carbon equivalent is higher ( $C_{eq} \approx 0.50$  %), then increase of metal cooling intensity results in rise of portion of martensite constituent in it and significant promotion of its susceptibility to cold crack formation.

Additional chromium alloying of highstrength steels, such as for example, 14Kh2GMR



**Figure 6.** Susceptibility of 06G2B (1), 09G2SYuch (2), 10KhSND (3), 12GN2MFAYu (4), 12GN3MFAYuDR (5), 14Kh2GMR (6) and 14KhGN2MDAFB (7) to cold crack formation depending on weld saturation by hydrogen and conditions of cooling of welded joints at  $w_{6/5} = 25$  (*a*) and 10 (*b*) °C/s

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and 14KhGN2MDAFB, promotes increase of  $C_{eq}$ in them up to 0.70 %, that has negative effect on delayed fracture of HAZ metal. Therefore, their welding requires simultaneous reduction of level of weld saturation by [H]<sub>dif</sub> and delay of  $w_{6/5}$ .

#### **Conclusions**

1. Probability of formation of longitudinal cold cracks in rigidly fixed welded joints from highstrength steels changes in wide ranges, however, they have specific dependencies related with effect of welding residual stresses on this process.

2. Probability of formation of cold cracks in welded joints from high-strength steels will be reduced to minimum at limitation of rate of their cooling up to  $w_{6/5} \le 10$  °C/s, content of diffusible hydrogen in the deposited metal up to 4 ml/100 g, level of residual stresses for steels with carbon equivalent  $C_{eq} = 0.35-0.45$  % up to  $0.9\sigma_{0.2}$ , and up to  $0.7\sigma_{0.2}$  and  $0.5\sigma_{0.2}$  with  $C_{eq} = 0.45-0.55$  and 0.60-0.70 %, respectively.

3. Increase of cooling rate of welded joint up to 25  $^{\circ}C/s$  and content of diffusible hydrogen in deposited metal up to 16 ml/100 g promotes the necessity of 1.7-1.9 times reduction of limits of residual stresses at C\_{eq} = 0.35–0.55 % and 2.5 times at C\_{eq} = 0.60–0.70 %.

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# MODELLING OF DYNAMIC CHARACTERISTICS OF A PULSED ARC WITH REFRACTORY CATHODE

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Self-consistent model of non-stationary processes of transfer of energy, pulse, mass and charge in the column and anode region of electric arc with refractory cathode was the basis to perform a detailed numerical analysis of dynamic characteristics for atmospheric-pressure argon arc with tungsten cathode and water copper-cooled anode at pulsed variation of electric current. An essential difference in dynamic behaviour of local and integral characteristics of arc plasma is shown, as well as the specifics of dynamics of thermal and electromagnetic processes running in the pulsed arc column and anode region. It is established that the velocities of transient processes in arc plasma at the pulse leading and trailing edges can also differ significantly. 28 Ref., 14 Figures.

**Keywords:** pulsed electric arc, refractory cathode, water-cooled anode, arc column, anode region, current pulse, dynamic characteristics, mathematical modelling

Nonconsumable electrode welding with pulsed modulation of arc current is becoming ever wider applied in modern welding fabrication, owing to additional possibilities of controlling the depth and shape of metal penetration, thermal cycle of welding, and, as a consequence, properties of the produced welded joint. These possibilities can be implemented through appropriate selection of the shape of welding pulses, duration and frequency of their repetition, value of base current and maximum pulse current. An important task, which is to be solved for theoretical substantiation of optimum modes of nonconsumable electrode welding with pulsed modulation of welding current, consists in numerical studies of non-stationary processes in the plasma of the column and near-electrode regions of the arc with the refractory cathode (primarily, anode processes, determining the interaction of arc plasma with the metal being welded) at pulsed variation of electric current.

There are a great number of publications devoted to theoretical study and mathematical modelling of processes of energy, mass and charge transfer in the column, near-electrode regions and in the electrodes of the arc with refractory cathode, in particular, in the metal being welded in nonconsumable-electrode inert-gas welding [1– 14]. However, the results presented in the majority of these publications pertain to stationary arcs running at direct current, except for [12– 14], which are devoted specifically to the processes of metal penetration in nonconsumableelectrode pulsed-arc welding. As regards the dynamic characteristics of pulsed arc with refractory cathode proper, note, for instance, works [15, 16], the first of which is devoted to experimental study of the above characteristics, whereas [16] gives the results of computational investigations of dynamic behaviour of both electric arc and weld pool in nonconsumable-electrode pulsed-arc spot welding. However, calculation data presented in this work do not allow analyzing the dynamic characteristics of pulsed arc at various rates of welding current variation at the leading and trailing edges of the pulse.

At arc running in pulsed mode it is possible to *a priori* single out two characteristic cases. If the rate of current variation is comparatively low, then non-stationary processes of transfer of energy, pulse, mass and charge in arc plasma run in the mode of a sequence of stationary states, each of which corresponds to the state of the stationary arc for instantaneous current value. Such a quasi-stationary arcing mode is realized, if the velocity of transient processes in the arc is significantly higher than the rate of current variation. In the second case, i.e. at high rates of arc current variation, the dynamic characteristics of arc plasma are the dominating factor. Computational investigation of these characteristics, as well as obtaining quantitative evaluations of current variation rates separating the quasi-stationary and non-stationary modes of running of pulsed argon arc with refractory (tungsten) cathode and copper water-cooled anode (Figure 1), is exactly the purpose of the work.

We will study the influence of pulsed variation of arc current on thermal, gas-dynamic and electromagnetic characteristics of its column, as





**Figure 1.** Schematic of design region for numerical modelling of pulsed arc with refractory cathode: 1 - tungsten cathode; 2 - nozzle for shielding gas feed; 3 - cathode region; 4 - arc column; 5 - anode region; 6 - copper water-cooled anode

well as on characteristics of its thermal and electric interaction with the anode surface separately for pulse leading and trailing edges (Figure 2). We will assume that arc current changes linearly both at leading and trailing edges of the pulse at the values of rise and fall times of current pulse b = 5, 20, 100 and 200 µs. We will also assume that having achieved its maximum (minimum) value, current remains constant during the time, sufficient for establishing of the respective stationary state of the arc.

Computational modelling of electric arc at the considered variation of current requires application of non-stationary mathematical model of the processes of energy, mass and charge transfer in arc plasma, which should incorporate the following interrelated models: model of thermal, electromagnetic and gas-dynamic processes in arc column plasma and models of near-electrode processes (see, for instance, [9, 16]). Model of anode processes is required for closing the model of non-stationary arc column by self-consistent boundary conditions on the anode, as well as for determination of the characteristics of thermal and electric interaction of such an arc with the anode surface [17]. As regards the model of cath-



Figure 2. Diagram of arc current variation at pulse leading and trailing edge

ode processes, then, as the theory of cathode phenomena, as well as processes in the near-cathode plasma of the electric arc with refractory cathode has been developed in sufficient detail [10, 18– 21], results of, for instance, [21], can be used as boundary conditions near the cathode.

At description of the processes in the plasma of the column of pulsed argon arc with tungsten cathode and copper water-cooled electrode we will use the model of isothermal ionization-equilibrium plasma [9, 22], and processes in near-anode plasma and on the anode surface will be described by the model of anode region, proposed in [17], as in the case of non-evaporating electrodes considered here, arc plasma can be regarded as single-component one, i.e. containing only the particles of shielding gas (argon). Let us use the data of [23] to determine the thermodynamic characteristics, transfer coefficients and radiation losses of such plasma, depending on its temperature and pressure. We will also assume that the considered system (see Figure 1) is axially symmetrical.

System of differential equations corresponding to the assumptions made, which describe the non-stationary thermal, gas-dynamic and electromagnetic processes in the arc column plasma, as well as dependencies of heat flow to the anode  $q_{\rm a}$  and anode potential drop  $U_{\rm a}$  =  $-\Delta \phi$  (where  $\Delta \phi$  is the difference of potentials between the outer boundary of the anode region and anode surface) on near-anode plasma temperature and electric current density in the anode, are given in [22]. It should be noted that at the frequencies of variation of arc electromagnetic characteristics  $\omega \leq 1.26 \cdot 10^6 \text{ s}^{-1}$ , considered in this work, which are determined by the rise and fall times of the current pulse, the skin-layer thickness [24] for arc plasma (atmospheric-pressure argon plasma at the temperature of 15,000 K) appears to be more than 3.3 cm, i.e. it is significantly higher than the characteristic dimensions of the arc. Therefore, application of Ohm's law and equation for scalar potential of electric field at description of non-stationary processes of charge transfer in arc plasma [22], i.e. neglecting bias currents, is quite justified.

We will define the design region, in which we will calculate the distributed characteristics of non-stationary arc plasma, as  $\Omega = \{0 < r < R; 0 < z < L\}$ , where *L* is the length of design region, actually equal to arc length, and *R* is the radius of design region, which is knowingly greater than the transverse dimensions of the arc (see Figure 1). As boundary conditions for the above equations, we will use the conditions on design region



boundaries, described in detail in [22], considering that we will interpret the boundary conditions for electromagnetic characteristics as those corresponding to the current value of arc current I(t), changing in time. As regards the initial conditions, we will assume that the distributed characteristics of arc plasma at moment of time t = 0 correspond to characteristics of the stationary arc at current, equal to the initial current value.

This boundary problem was solved numerically, by the finite difference method. Gas-dynamics and convective heat transfer equations were solved using joint Euler–Lagrange method [25, 26], adapted to the conditions of compressible medium. During calculations model parameters were selected as follows. Dimensions of design region were L = 3 mm, R = 8 mm; net parameters were time step  $\tau = 0.5 \ \mu s$ ; net steps by spatial coordinates were  $h_r = 0.125$  mm,  $h_z =$ = 0.06 mm. Ambient temperature was taken to be equal to 500 K, and temperature of the surface of copper water-cooled anode was  $T_s = 720$  K [22]. Value of radius in the region of cathode binding  $R_c$  (see Figure 1) was determined on the basis of recommendations of [21] so that the maximum value of density of electric current in this region was constant ( $j_{c0} = 10^8 \text{ A/m}^2$ ) in the entire studied range of arc current variation (50– 200 A). Maximum plasma temperature near the cathode was also selected to be constant ( $T_{c0}$  = = 20,500 K) [21].

Dynamics of variation of temperature field and pattern of arc plasma flow at fast variation of arc current ( $b = 5 \mu m$ ) is shown in Figures 3 and 4 (time in these Figures is calculated from the moment of start of current variation). In Figures 3, 4 the isotherms correspond to the respective temperatures 1; 2; 4; 6; 8; 10; 12; 14; 16; 18 kK from the arc periphery to its axis. Calculation results are quite predictable: at pulse leading edge (at current increase from 50 up to 200 A) the bell-shaped isothermal lines in the arc column become wider; contrarily, at the trailing edge (at current decrease from 200 to 50 A) high-temperature current-conducting region of arc plasma is contracted. In both the cases, a certain time (about 100 µs at increase of arc current and about 120 µs at its decrease) is required for the temperature field and pattern of arc plasma flowing to reach the respective stationary states.

Unlike the above-given general pattern of arcing dynamics, variation in time of individual local and integral characteristics of the column and anode region of the arc with refractory cathode at pulsed variation of electric current has a number of specific features. We will select the following arc column characteristics as those, the variation dynamics of which we will analyze further:  $T_{\rm col}$  and  $j_{\rm col}$  are the plasma temperature and electric current density on the column axis, calculated in its middle section (at z = 1.5 mm);  $R_{\rm col}$  is the characteristic radius of current-conducting plasma region in the same section, defined as circle radius, within which 95 % of instantaneous value of arc current is concentrated.

Figures 5–7 give the variation in time (time is calculated since the moment of the start of current variation) of the above characteristics for pulse leading and trailing edges at b = 20, 100 and 200 µs (solid, dashed and dotted lines, respectively).

As follows from calculated dependencies presented in these Figures, plasma temperature in the center of arc column is its characteristic with the least inertia. This accounts for practically instantaneous, proportional to current variation (at least, at  $b \ge 20 \ \mu s$ ), change of the efficiency of Joule heat sources, leading to the respective increase or lowering of  $T_{col}$  (see Figure 5). A slight maximum of arc plasma temperature observed at the leading edge of current pulse at b == 20  $\mu$ s is related to its heating by rising current (see solid curve in Figure 6, a) to temperatures, exceeding  $T_{\rm col}$  value for a stationary 200 A arc, and to subsequent cooling down due to slower convective cooling (characteristic time of relaxation of arc column plasma temperature under the considered conditions is equal to about 30  $\mu$ s). With increase of pulse rise time up to 100  $\mu$ s and higher, this maximum practically disappears, as the rate of convective cooling, determined by inertia of gas-dynamic processes in arc plasma, becomes commensurate with the rate of increase of arc current, and, therefore, also of Joule heating of plasma at b increase (see dashed and dotted curves in Figure 5, a). It should be noted that such an effect is practically not manifested at current drop at pulse trailing edge (see Figure 5, b).

As regards current density in the center of arc column, since  $j_{col}$  is the product of plasma electric conductivity determined by its temperature value in the same point, by electric field intensity, determined by distribution of temperature (electric conductivity) across the entire column section, the above characteristic has somewhat greater inertia than  $T_{col}$ . Local maximum of  $j_{col}$  observed at the leading edge of current pulse at  $b = 20 \ \mu s$  turns out to be more pronounced (current density in the center of pulsed arc column at the moment, when its current reaches 200 A, is by almost 25 %



higher than the respective value for a stationary 200 A arc), and subsequent lowering of  $j_{col}$  and establishing of its stationary value occurs during the time of about 50 µs (see solid curve in Figure 6, a). At lowering of the rate of current rise in the pulse (b = 100 and 200 µs), this maximum, similar to temperature maximum, becomes ever less pronounced (see hatched and dotted curves in Figure 6, a). Unlike  $T_{col}$  behaviour at the pulse trailing edge, electric current density in the arc column at total current drop has a local minimum, the absolute value of which decreases at bincrease (see Figure 6, b). The features of variation of density of electric current in arc plasma noted here are in many respects characteristic also for variation in time of voltage in pulsed arc column. In particular, the difference in the time of transient processes in the arc at the pulse leading and trailing edges at low values of fall and rise times of pulse current is one of the causes for formation of a hysteresis loop on pulsed arc volt-ampere characteristic [15, 27].

Arc column characteristic with the highest inertia is the radius of its current-conducting region that is attributable to restructuring of temperature filed over the entire column cross-section, necessary for  $R_{\rm col}$  variation. Characteristic time for establishment of stationary value of this radius, after the arc has reached its stationary (maximum) value in the case  $b = 20 \ \mu s$ , is equal to approximately 100 µs (see solid curve in Figure 7, a). It should be noted that the characteristic time of  $R_{col}$  variation at arc current drop is essentially smaller, and is equal to about  $60 \ \mu s$ at  $b = 20 \ \mu s$  (see solid curve in Figure 7, b). Finally, the time of establishment of a stationary value of radius of arc column current-conducting region decreases significantly at b increase due to the fact that  $R_{\rm col}$  variation partially occurs already during current rise or drop (see dashed and dotted lines in Figure 7).

Non-stationary processes occurring in the arc anode region, are illustrated by graphs (Figures 8–13) of variation in time of both the local characteristics of anode processes:  $T_{a0}$  is the axial value of plasma temperature near anode surface (at z = 3 mm),  $j_{a0}$  and  $q_{a0}$  are the density of electric current on the anode and density of heat flow to the anode determined in the center of the region of anode binding of the arc, and of the integral characteristics of the above processes:  $P_a$  is the total thermal flow to the anode;  $R_a$  and  $R_h$  are the radii of current channel and region of thermal impact of the arc on anode surface ( $R_a$ and  $R_h$  are understood to be the radii of circumferences on anode surface, within which 95 % of current values of total arc current I(t) and total heat flow to the anode  $P_{\rm a}(t)$  are concentrated, respectively). Solid, dashed and dotted curves in the above Figures correspond to b = 20, 100 and 200 µs.

Regularities of dynamic variation of local and integral characteristics of the anode region of the arc with refractory cathode and copper watercooled anode at application of electric current pulse, are not trivial and require detailed physical interpretation. So, for instance, at high rate of arc current variation ( $b = 20 \ \mu s$ ), instead of the anticipated increase of axial value of near-anode plasma temperature at the pulse leading edge, and its decrease at the trailing edge, respectively, first a certain lowering of  $T_{a0}$  at the leading edge and its more noticeable increase at the pulse trailing edge is observed (see solid curves in Figure 8). This effect is largely related to the features of thermal state dynamics and pattern of plasma flowing in the arc column at pulsed variation of current. To analyze this effect, we will consider the condition of local heat balance in the anode region [22]:

$$q_{\chi} + q_j = \Delta \varphi j_{\rm a} + q_{\rm a}. \tag{1}$$

Here,  $q_{\chi} = -\chi \frac{\partial T}{\partial z} \Big|_{z = L}$  is the thermal flow from the arc column plasma, where  $\chi$  is the co-

efficient of heat conductivity of arc plasma;  $q_j = j_a \frac{k}{e} \left(\frac{5}{2} - \delta\right) T \Big|_{z=L}$  is the flow of energy brought to the anode region by column plasma electrons, where  $j_a = -j_{z}\Big|_{z=L}$  is the density of electric current in near-anode plasma; k is the Boltzmann constant; e is the electron charge;  $\delta$ is the constant of electron thermal diffusion;  $\Delta \varphi j_a$ are the energy losses for maintaining the anode layer, while values  $\Delta \varphi$  and  $q_a$  are determined on the basis of the model of anode processes [17], depending on near-anode plasma temperature  $T_a$  $= T|_{j=L}$ , anode surface temperature  $T_s$  and current density in the anode region  $j_a$ .

As was already noted, gas-dynamic processes have the greatest inertia in the arc column. At the beginning of the current pulse trailing edge the maximum velocity of plasma motion along the arc column axis is equal to almost 330 m/s (see Figure 4). Despite the fast drop of current at  $b = 20 \ \mu s$  and respective lowering of volume density of electromagnetic force, the plasma, moving by inertia, continues transporting thermal energy from arc column towards the anode by convective flows for a certain time, thus ensuring the highest  $q_{\gamma}$  values. At the same time,





**Figure 3.** Dynamics of temperature fields and plasma velocity in pulsed arc column at current rise from 50 up to 200 A ( $b = 5 \ \mu s$ ): a - t = 0 ( $V_{max} = 120.1 \ m/s$ );  $b - t = 50 \ \mu s$  ( $V_{max} = 328.7 \ m/s$ );  $c - t = 100 \ \mu s$  ( $V_{max} = 329.2 \ m/s$ )

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**Figure 4.** Dynamics of temperature fields and plasma velocities in pulsed arc column at current decrease from 200 to 50 A ( $b = 5 \mu m$ ): a - t = 0 ( $V_{max} = 329.4 \text{ m/s}$ );  $b - t = 60 \mu s$  ( $V_{max} = 120 \text{ m/s}$ ),  $c - t = 120 \mu s$  ( $V_{max} = 120.7 \text{ m/s}$ )

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Figure 5. Change of plasma temperature in the center of arc column at leading (a) and trailing (b) edges of current pulse



Figure 6. Variation of electric current density in the center of arc column at pulse leading (a) and trailing (b) edges







Figure 8. Variation of axial value of near-anode plasma temperature at pulse leading (a) and trailing (b) edges

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Figure 9. Variation of axial value of electric current density in the anode region at the pulse leading (a) and trailing (b) edges

at lowering of current density in the anode region (see solid curve in Figure 9, b) heat flow density due to energy transfer by charged particles, i.e.  $q_i$  value, decreases. The addends in the right-hand part of energy balance (1) also decrease at lowering of total arc current, in view of reduction of current density and density of heat flow on the anode (see solid curves in Figure 9, b and 10, b). With such a tendency of variation of heat balance components, the heat flow due to heat conductivity, has the dominating role in the initial period of current variation, leading exactly to local increase of  $T_{a0}$ . Later on, when the intensity of gas-dynamic flows drops, the temperature of near-anode plasma in the center of the region of anode binding of the arc starts decreasing monotonically to values, corresponding to stationary arc at the current of 50 A (characteristic value of relaxation time at the pulse trailing edge is equal to about 50  $\mu$ s). A reverse situation is found at the pulse leading edge, the characteristic time of temperature relaxation being essentially lower and equal to a value of the order of 20  $\mu$ s. The described effect is not observed at  $b \ge 100 \ \mu s$  (see dashed and dotted curves in Figure 8), as with such pulse rise and fall times, the pattern of motion of arc column plasma has

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enough time for restructuring during current variation.

The non-stationarity effects are manifested to the greatest degree in the dynamics of variation in time of the density of electric current and density of heat flow on the anode in the center of the region of anode binding of the arc (see Figures 9 and 10). The main feature of these dependencies is their non-monotonic nature with formation of local maximums (at the pulse leading edge) and minimums (at the trailing edge), which are reached by the moment of time, corresponding to the end of rise or drop of arc current. In particular, at a high rate of total current increase from 50 up to 200 A ( $b = 20 \ \mu s$ ), maximum current density in the axial zone of the anode region is more than 2 times higher than the respective value for the stationary arc at I == 200 A, and the characteristic relaxation time of  $j_{a0}$  is equal to about 80 µs (see solid curve in Figure 9, a). At the trailing edge in the minimum point the axial value of current density on the anode turns out to be almost 1.5 times lower than for stationary 50 A arc at somewhat longer relaxation time, equal to about 100 µs (see solid curve in Figure 9, b).



**Figure 10.** Change of axial value of density of heat flow to the anode at pulse leading (*a*) and trailing (*b*) edges (markers show  $q_{a0}$  values for stationary arc at respective current values:  $\Delta - b = 20$ ;  $O = 200 \ \mu s$ )



**Figure 11.** Variation of total heat flow to the anode at pulse leading (*a*) and trailing (*b*) edges (markers show  $P_a$  values for stationary arc at respective current values:  $\Delta - b = 20$ ;  $O = 200 \ \mu s$ )

Let us consider the cause for such an extreme change of current density at the pulse leading edge at  $b = 20 \ \mu s$ , when this effect is manifested to the greatest degree. Let us bear in mind that in this case the rate of arc current variation is essentially higher than the rates of relaxation of gas-dynamic and thermal processes in arc plasma. Moreover, as shown by calculations, the radius of current-conducting region on the anode at total current rise first decreases markedly and only then it starts growing, reaching its steady-state value, corresponding to 200 A arc, during the time of approximately 100 µs (see solid curve in Figure 12, a). All that leads to the situation when at the rising arc current its density in the center of the anode binding region first rises abruptly, and then smoothly decreases, as shown in Figure 9, a.

At current pulse trailing edge at  $b = 20 \ \mu$ s the radius of current-conducting region on the anode shows an even more nontrivial behaviour, namely: value  $R_a$  during arc current drop decreases somewhat, and then rises and only later it drops again to values, characteristic for the stationary arc at the current of 50 A (see solid curve in Figure 12, b). The result of such a behaviour of the radius of current-conducting region on the anode is the fact that  $j_{a0}$  minimum turns out to be less pronounced (see solid curve in Figure 9, b). Extreme nature of  $j_{a0}(t)$  variation is manifested, even though to a smaller degree, also at lower rates of current variation, i.e. at b = 100 and 200 µs (see dashed and dotted curves in Figure 9).

As the density of heat flow to the anode at other conditions being equal, is practically proportional to current density on the anode, dynamics of  $q_{\rm a0}$  variation is, on the whole, similar to that of variation of axial value of electric current density in the anode region (see Figures 9 and 10). Axial values of density of heat flow to the anode for a stationary arc at respective values of total current are indicative of the fact that in the case of  $b = 20 \ \mu s$ , local characteristics of the arc anode region are essentially non-stationary, whereas in the case of  $b = 200 \ \mu s$  the change of the above characteristics at current variation takes place practically in the quasi-stationary mode, i.e. running of a pulsed arc at  $b = 200 \ \mu s$ is a sequence of the states of a stationary arc, running at the respective current values. Thus, a value of about 100 µs can be selected as the characteristic time of variation of pulsed arc current (pulse rise and fall times), separating the non-stationary and quasi-stationary modes of arcing in terms of local characteristics of electric and thermal impact on the anode. It should be noted that the extreme nature of variation of local electric and thermal characteristics of anode region of pulsed arc with refractory cathode can



Figure 12. Variation of radius of arc current channel on the anode at pulse leading (a) and trailing (b) edges



**Figure 13.** Variation of radius of the region of arc thermal impact on the anode on pulse leading (*a*) and trailing (*b*) edges

lead to an important technological result of nonconsumable electrode pulsed-arc welding - an essential increase of arc penetrability due to contraction of its electric and thermal impact on weld pool surface and resulting intensification of the processes of heat transfer in its volume.

A characteristic of arc anode region, the most sensitive to the rate of electric current variation, is its such an integral characteristic as total heat power, applied to the anode (see Figure 11). Despite the fact that at low b values the specific heat flow to the anode is essentially non-stationary (compare solid curves and respective markers in Figure 10),  $P_{\rm a}$  value changes in an almost quasi-stationary manner (compare solid curves and respective markers in Figure 11). At larger values of pulse rise and fall times ( $b = 200 \ \mu s$ ) values of power applied to the anode by the stationary arc at respective current values practically coincide with values determined by  $P_a(t)$ dependence for the pulsed arc (compare dotted curves and respective markers in Figure 11).



**Figure 14.** Variations of the degree of ionization non-equilibrium of arc plasma in arc column center at current pulse leading edge at b = 20, 100 and 200 µs (solid, dashed and dotted curves, respectively)

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Results of numerical modelling of dynamic characteristics of pulsed arc with refractory cathode and copper water-cooled anode are indicative of the fact that in the studied range of current pulse rise and fall times the characteristic time of variation of arc plasma thermal state can be equal to  $10^{-5}-10^{-4}$  s. As these values are commensurate with the characteristic times of ionization-recombination processes in atmosphericpressure argon plasma [28], it is necessary to evaluate the appropriateness of application of the model of ionization-equilibrium plasma and calculated on its basis temperature dependencies of thermodynamic characteristics, transfer coefficients and radiation losses of such plasma. With this purpose we will introduce parameter  $\gamma =$  $= |\alpha - \alpha_{\rm e}| / \alpha_{\rm e}$ , characterizing ionization nonequilibrium of arc column plasma, where  $\alpha$  is the degree of plasma ionization calculated allowing for the final rates of ionization-recombination processes, and  $\alpha_e$  is its equilibrium value, calculated using Saha equations. Figure 14 shows the change in time of  $\gamma$  parameter for pulsed arc column plasma at  $T_{\rm col}$  variation, according to dependencies given in Figure 5, a. As follows from calculation data given in Figure 14, degree of ionization non-equilibrium of arc column plasma under the considered conditions does not exceed 1.5 % that allows regarding application of the model of ionization-equilibrium plasma as quite justified.

On the whole, regularities of dynamic behaviour of local and integral characteristics of the column and anode region of pulsed arc with tungsten cathode and copper water-cooled anode, described in this paper, lead to the following conclusions.

1. Running of the arc with refractory cathode in the pulsed-periodic mode is accompanied by



an essential variation of electromagnetic, thermal and gas-dynamic characteristics of arc plasma, as well as its electric and thermal impact on the anode surface. Dynamic behaviour of the above characteristics largely depends on the rate of arc current variation at pulse edges and is different for the leading and trailing edges. Gas-dynamic processes are the link with the highest inertia in the process of restructuring of electromagnetic fields, thermal state and pattern of arc plasma flowing at variation of arc current.

2. At great steepness of pulse edges (more than  $5 \cdot 10^6$  A/s rate of current variation) change of characteristics of the column and anode region of pulsed arc occurs in two stages: stage of arc current variation and stage of transient processes. At increase (decrease) of current, heat flow density and anode current density can be 2 times greater (1.5 times smaller) than the respective values characteristic for the direct current arc, at current equal to arc current in the pulse (pause). At transient process stage, relaxation of thermal and gas-dynamic state of arc plasma to values characteristic for the stationary arc at the respective current value, takes place. Durations of relaxation processes depend on the value of base current and pulse current, and can differ essentially for the local and integral characteristics of plasma of arc column and anode region.

3. At current variation at pulse edges with at the rate below  $10^6 \mbox{ A/s}$  (pulse rise and fall times of more than 100  $\mu$ s), the processes related to current variation and relaxation processes occur simultaneously, as a result of which the non-stationary process of pulsed arc running is realized as a sequence of states characteristic for the stationary arc at the respective values of current (quasi-stationary mode).

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# MAGNETRON-SPUTTERED NANOCOMPOSITE nc-TiC/a-C COATINGS

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One of the main directions in current progress of surface engineering is development of a nanocomposite structure, with at least one phase with less than 100 nm size of structural element among its components. Presence of a multiphase structure with dissimilar grain boundaries creates a barrier for their coarsening, thus providing stability of the formed coating structure. This work is devoted to investigation of the process of formation of a nanocomposite nc-TiC/a-C coating on substrates from 08Kh18N10T, Kh12M steels and VT1-0 titanium by the method of magnetron sputtering of graphite and titanium targets. In order to control the coating composition, a design procedure was developed, which envisages a change of power of magnetron discharge with titanium target at constant power of discharge with graphite target, that would allow producing coatings in the range of compositions of 42.5-70 at.% C and 57.5-30 at.% Ti. Coatings were studied by the methods of X-ray diffraction, Raman spectroscopy, X-ray photoelectron spectroscopy and microindentation. It is established that nanocrystalline TiC phase takes up 80 % and amorphous carbon matrix is 20 % of the coating structure. It is found that the degree of carbon ordering depends on coating composition. It is shown that the size of TiC grain and coating hardness depend on Ti/C ratio. Minimum size of TiC grain (2.9–4.3 nm) and maximum hardness (up to 30–38 GPa) are achieved at Ti/C ratio (in at.%) of 46/54. Maximum normalized hardness  $H/E^* = 0.134$ , which is the characteristic of coating material resistance to plastic deformation, is achieved on the substrate of 08Kh10N10T steel. 19 Ref., 3 Tables, 7 Figures.

**Keywords:** nanocomposite coating, magnetron sputtering, titanium carbide, amorphous carbon, grain size, structure, hardness

One of the main directions in current progress of materials science and surface engineering is use of nanostructural state of consolidated materials and coatings. This is due to the possibility of realization of dimensional effects, arising at refinement of structure grain to 100 nm and less, and accompanied by an essential change of mechanical, thermal, kinetic, electrical, magnetic and optical characteristics of materials [1–4].

However, practical application of materials and coatings with a totally nanosized structure is confronted with the problem of stability of such a structure, grain growth caused both by natural ageing and by the impact of external factors, namely mechanical, thermal, radiation, etc. This leads to the material loosing the specific properties, inherent to the nanocrystalline structural state.

Therefore, another direction of application of the advantages of nanostructural state of materials and coatings, being actively developed now, is creation of a nanocomposite structure, which has at least one phase with structural element size of less than 100 nm among its components [4]. Multiphase nature of the structure with dissimilar grain boundaries prevents grain coarsening that ensures the stability of the formed coating structure. It is established that owing to specifics of their structure, nanocomposites are characterized by improved physico-mechanical and catalytic properties [4].

Among nanocomposite coatings applied by vacuum deposition, a special place belongs to the group of coatings produced by the method of magnetron sputtering. Their structure consists of practically noninteracting phases with average size of structural elements below 100 nm [5–8]. Such structural elements are the amorphous matrix and nanocrystalline phase inclusions. Structural schematic of such a nanocomposite structure, which is synthesized by deposition of flows from several targets on the substrate in the atmosphere of reactive gases, is shown in Figure 1.

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The best studied are such systems of nanocomposite structures as Me–C, Me–Si–N, where the amorphous matrix is carbon or  $Si_3N_4$  with solid inclusions, mainly of carbides and nitrides of titanium, zirconium and chromium [9–11].

A number of research efforts were devoted to studying the influence of the composition of magnetron nc-TiC/a-C coating on nanosized nature of their structure, mechanical and tribotechnical properties [11-14]. In [13], at production of nc-TiC/a-C coating by magnetron sputtering of targets from compressed TiC and C mixtures in the proportion of 70/30, 50/50 and 30/70 (in mol.%), the coating, deposited by sputtering of 50/50 target, is a nanocomposite one at TiC grain size in the range of 6–11 nm and has hardness HV0.05-4190. When 30/70 target is used, coating structure is completely X-ray amorphous, and the authors classified it as metal-containing coating from amorphous carbon. Measurement of tribotechnical properties showed that TiC/Ccoating produced from 30/70 TiC/C target has the best characteristics. Friction coefficient under dry friction conditions on 100St6 steel was equal to 0.08-0.14 (at 0.4-0.5 for other coating compositions).

Evaluation of the influence of a-C amorphous phase content in the ranges from 0 to 100 % on the structure and mechanical properties of TiC/C coatings, produced at magnetron sputtering of titanium and graphite targets, demonstrated an increase of coating hardness from 17 up to 22 GPa at variation of its amount from 0 up to 45 %. At further increase of a-C-phase content lowering of hardness is observed, but coating tribological characteristics are improved. The best result was achieved at 65 % content, when the friction coefficient was below 0.2, wear rate was about  $10^{-7}$  mm<sup>3</sup>/(N·m) at coating hardness of 10–15 GPa [14].

Work [15] is a study of the influence of the composition of magnetron-sputtered TiC/a-C, WC/a-C and TiBC/a-C nanocomposite coatings on their structure and tribotechnical properties. In the case of TiC/a-C, the composition was varied within the range of content (in at.%) of 43.6-85.5 C; at WC/a-C it was 33.9-72.5, and at TiBC/a-C it was 31.2-71.3 C. A refinement of hard inclusion grain was observed in carbide-containing systems (from 30 to 2–3 nm for TiC, and from 9 to 2-3 nm for WC). TiBC grain in the case of 49.2-71.3 at.% C had the size of 2-3 nm. TiC/a-C, and WC/a-C nanocomposite coatings improve their tribomechanical properties at increase of carbon content with lowering of friction coefficient (from 0.31 to 0.04



Figure 1. Schematic of nanocomposite structure

for TiC/a-C, and from 0.84 to 0.19 for WC/a-C) and of the degree of wear, with lowering of coating hardness (from 22.3 to 8 GPa for TiC/a-C, and from 35.7 to 15.8 GPa for WC/a-C).

Similar results were presented in [16] and [17]. In [16] formation of X-ray amorphous phase at not less than 8 at.% Ti in the coating and of nanocomposite structure at 16 at.% Ti was confirmed. Coarsening of TiC grain from 5 up to 16 nm in the range of TiC content of 16–48 at.% and achievement of maximum hardness of 31 GPa at 30 at.% TiC was noted.

Work [17] is a study of the influence of carbon content on mechanical characteristics of nc-TiC/a-C coating in the range of 55–95 at.% C. It is shown that at content in the ranges of 70– 95 at.% C, the index of normalized hardness  $H/E^*$  ( $E^*$  is the contact modulus of elasticity) was equal to 0.10–0.15, rising with increase of carbon content (%).

This work was devoted to investigation of the process of formation of nanocomposite TiC/C coating under the conditions of magnetron sputtering of separate targets from graphite and titanium, studying the influence of coating composition on its structure and mechanical properties.

**Experimental and investigation procedures.** The coating was deposited in an upgraded vacuum system VU-1BS, which was fitted with DC magnetron sputtering module consisting of two magnetrons: Magnetron 1 with a disc target (88 mm diameter, 4 mm thickness) from MPG-7 graphite of 99.98 % purity, and Magnetron 2 with a rectangular target (90  $\times$  58  $\times$  4 mm) from VT1-0 titanium (Figure 2). Magnetrons are mounted on one flange so that the angle between the target surfaces was equal to 150°. This enabled simultaneous or alternating deposition of coatings on a stationary substrate from two mag-





**Figure 2.** Module of magnetron sputtering, consisting of Magnetron 1 (1) and Magnetron 2 (2)

netrons with the same substrate-to-target distance equal to 110 mm.

Samples from 08Kh18N10T steel and VT1-0 titanium of  $65 \times 30 \times 0.5$  mm size, and Kh12M steel samples of 25 mm diameter and 6 mm thickness were used.

Before placing into the vacuum chamber, the sample was cleaned in an ultrasonic bath, successively filled with acetone and ethyl alcohol. In the vacuum at the pressure of  $5 \cdot 10^{-4}$  Pa the sample was heated at 150 °C for 20 min, then without switching off the heater, sample surface was cleaned by argon ion bombardment in a glowing DC discharge at 1.3 Pa, 1100 V for 20 min. The coating was deposited in argon at working pressure of 0.4 Pa.

The process of forming TiC/C coating on the sample surface (with roughness  $Ra = 0.045 \ \mu\text{m}$ ) consisted of two stages: deposition of an bond coat of titanium of 0.26–0.13  $\mu$ m thickness with Magnetron 2 in the mode with negative voltage bias on the substrate at bias potential  $U_{\rm b} = -100 \ \text{V}$ , and deposition of TiC/C coating ( $\delta = 1.5$ –3.0  $\mu$ m) using two magnetrons at  $U_{\rm b} = 0 \ \text{V}$ . Here, carbon was deposited at constant values of specific power of magnetron discharge  $\Delta P_1 = 8.5 \ \text{W/cm}^2$  and carbon deposition rate  $V_{\rm C} = 0.6 \ \mu\text{m}/\text{h}$ . Value of discharge power  $P_2$  required to produce the specified percentage of titanium content in the coating was set on Magnetron 2.

To assess the power of Magnetron 2, required to produce the specified titanium content in TiC/C coating, a calculation procedure was developed, which consisted of the following stages:

1. Titanium content in the coating (at.%)  $C_{\text{Ti}}^{\text{at}}$ , proceeding from the known deposition rates of both the coating components (titanium and carbon), will be equal to

$$C_{\rm Ti}^{\rm at} = \frac{V_{\rm Ti}}{V_{\rm Ti} + V_{\rm C}} 100;$$
 (1)

2. Experimental determination of velocity of carbon coating atom deposition (C at/h)  $V_{\rm C}$  with Magnetron 1 on the surface of reference sample across coating thickness, measured with a profile measurement device:

$$V_{\rm C} = \frac{Q_{\rm C}N}{A_{\rm C}}, \ {\rm C} \ {\rm at./h}, \tag{2}$$

where  $Q_{\rm C} = \delta_{\rm c} \rho_{\rm c} s$  is the carbon deposition rate, g/h ( $\delta_{\rm c} = 0.6 \cdot 10^{-4} \text{ cm/h} - \text{ coating thickness}$ ;  $\rho_{\rm c} = 2.2 \text{ g/cm}^3 - \text{ electrode density}$ ;  $s = 14.5 \text{ cm}^2 - \text{ sample area}$ ;  $N = 6.02 \cdot 10^{23} \text{ at/mole is the Avogadro number}$ ;  $A_{\rm C} = 12.01$  is the carbon atomic weight;

$$V_{\rm C} = \frac{\delta_{\rm c} \rho_{\rm c} N s}{A_{\rm C}} = 0.94 \cdot 10^{20}, \ {\rm C} \ {\rm at/h}.$$
 (3)

3. Experimental determination of the rate of titanium deposition at nominal power of Magnetron 2, equal to 325 W. Similar to previous calculation procedure

$$V_{\rm Ti}^{\rm nom} = \frac{\delta_{\rm Ti} \rho_{\rm Ti} Ns}{A_{\rm Ti}} = 1.27 \cdot 10^{20}, \, {\rm Ti} \, {\rm at/h},$$
 (4)

where  $\delta_{\text{Ti}} = 1.56 \cdot 10^{-4} \text{ cm}$ ;  $\rho_{\text{Ti}} = 4.5 \text{ g/cm}^3$ ,  $A_{\text{Ti}} = 47.9$ .

4. As at magnetron sputtering the coating deposition rate is directly proportional to magnetron discharge power, at constant  $V_{\rm C}$  the required  $V_{\rm Ti}^{\rm w}$  depends on working power of Magnetron 2  $P_2^{\rm w}$ , referred to the value of nominal power 325 W, where  $V_{\rm Ti}^{\rm nom}$  is known, i.e.  $V_{\rm Ti}^{\rm w}/V_{\rm Ti}^{\rm nom} = P_2^{\rm w}/P_2^{\rm nom}$ ;

$$P_{2}^{w} = P_{2}^{nom} \frac{V_{Ti}^{w}}{V_{Ti}^{nom}} = V_{Ti}^{w} \frac{P_{2}^{nom}}{V_{Ti}^{nom}}, W.$$
 (5)

Proceeding from expressions (1) and (5)

$$V_{\rm Ti}^{\rm w} = \frac{C_{\rm Ti}^{\rm at} V_{\rm C}}{100 - C_{\rm Ti}^{\rm at}}, \quad P_2^{\rm w} = \frac{P_2^{\rm nom}}{V_{\rm Ti}^{\rm nom}} \frac{C_{\rm Ti}^{\rm at} V_{\rm C}}{100 - C_{\rm Ti}^{\rm at}}.$$
 (6)

Using available data on  $V_{\rm C}$  and  $V_{\rm Ti}^{\rm nom},$  we obtain:

$$P_2^{\rm w} = 240.5 \ \frac{C_{\rm Ti}^{\rm at}}{100 - C_{\rm Ti}^{\rm at}}, {\rm W}.$$
 (7)



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Figure 3 gives the graphic image of the obtained connection of titanium content in the coating with working power of Magnetron 2.

Phase analysis of coatings was conducted by the method of X-ray diffraction, using X-ray diffractometer Philips X'Pert — MPD with Cu $K_{\alpha}$ X-ray source (wave length  $\lambda = 0.15418$  nm). Xray diffractograms (XD) were taken in sliding geometry (2 $\theta$  scanning): angle of incident beam did not change and was equal to 4° relative to sample surface; full angular range of diffraction spectrum recording by  $2\theta = 20-80^\circ$ , minimum step of 0.02°. Proceeding from analysis of XD spectra, the following parameters were determined: coating phase composition; average size of TiC phase grains (D); magnitude of average strain in TiC layer ( $\epsilon$ ).

Method of Raman spectroscopy was used for determination of configurations of carbon chemical bonds in the coating. Spectra of Raman scattering (RS) were measured in reflection geometry at room temperature, using triple Raman spectrometer T-64000 Horiba Jobin-Yvon, fitted with cooled CCD detector. Ar-Kr line of an ion laser with 488 nm wave length was used for excitation. Radiation was focused on the sample into a spot of 1  $\mu$ m size, and power radiation hitting the sample was about 1 mW.

Ratios of carbon and TiC phases in TiC/C coating were determined by the method of X-ray photoelectron spectroscopy, using a high resolution electron spectrometer RIBER LAS 2000.

Determination of mechanical characteristics of coatings was conducted by the method of microindentation, using Micron-Gamma indentor [18]. Values of the characteristics were calculated automatically to ISO 14577-1:2002 standard.

**Results and their discussion.** Main characteristics of nc-TiC/a-C coatings of various composition deposited on substrates from 08Kh18N10T steels, as well as parameters of their deposition process, are given in Table 1.

As is seen from Table 1, in order to increase carbon content in TiC coating from 42.5 to 70 at.%, titanium deposition rate was reduced 3 times (at  $V_{\rm C} = 0.6 \ \mu {\rm m/h}$ ). With increase of carbon content in the coating up to 54 at.% its hardness rises. For instance, 8TiC in (42.5 at.% C) and 19TiC (54 at.% C) samples hardness is equal to 7.7 and 17.5 GPa, respectively. At higher carbon content coating hardness decreases. Thus, changing the velocity of titanium target sputtering at preservation of optimum rate of carbon deposition allows controlling the composition and hardness value of the produced TiC/C coating.



Figure 3. Dependence of required power of Magnetron 2 on required titanium content in  $\rm TiC/C$  coating

Figure 4 gives XD of 8TiC, 19TiC and 26TiC samples with the respective Tables, which show lattice parameters, interplanar spacings, as well as crystal dimensions for each reflex.

Analysis of XD spectra shows that in the case of 8TiC and 19TiC samples a polycrystalline nanostructure with TiC crystal dimensions of 10.2-5.3 and 4.3-2.9 nm, respectively, forms in TiC/C coatings. With increase of carbon content in the coating, nanocrystal dimensions become smaller, and in 26TiC sample (70 at.% C) TiC phase is not found at all. As (111), (200), (220) and (311) reflexes are visible in the spectra, this is indicative of absence of any privileged direction of crystal growth.

Nanocrystalline structure of TiC/C coating ( $\delta = 3 \mu m$ ) studied by SEM at ×10,000 and ×100,000 magnification is shown in Figure 5, where TiC grains of 10–30 nm size are observed.

Structure of nc-TiC/a-C coatings is revealed by the spectrum of photoelectrons, generated under the impact of monochromatic X-ray radiation on the coating surface (Figure 6).

The spectrum has a wider carbon peak C1s, which is decomposed into two peaks of electrons with bond energies of 284.4(1) and 281.2(2) eV, corresponding to the presence of carbon phases and TiC in the coating. The ratio of values of peaks (1) and (2) shows that the phase of

Table 1. Parameters of magnetron deposition process and characteristics of nanocomposite nc-TiC/a-C coatings

Sample	Calculated coating composition, at.%	Titanium deposition rate, μm/h	TiC deposition rate, µm∕h	Coating thickness δ, μm	Hardness <i>H</i> , GPa
8TiC	42.5 C 57.5 Ti	1.56	2.16	2.4	7.7
14TiC	50 C 50 Ti	1.05	1.65	1.9	8.0
19TiC	54 C 46 Ti	0.81	1.40	1.8	17.5
26TiC	70 C 30 Ti	0.50	1.10	1.2	15.7



8TiC						
Reflexes	Interplanar space <i>d</i> , nm	Lattice constant, nm	Crystallite size, nm			
(111)	0.2518	0.436	10.2			
(200)	0.2180	0.436	8.1			
(220)	0.1539	0.435	6.2			
(311)	0.1312	0.435	5.3			

19TiC						
Reflexes	Interplanar space <i>d</i> , nm	Lattice constant, nm	Crystallite size, nm			
(111)	0.2519	0.436	3.4			
(200)	0.2183	0.436	4.3			
(220)	0.1535	0.434	2.9			

**Figure 4.** X-ray diffractograms of TiC/C coatings of various composition, at.%: a - 8TiC (42.5 C; 57.5 Ti); b - 19TiC (54 C; 46 Ti); c - 26TiC (70 C; 30 Ti)



Figure 5. Microstructure of nc-TiC/a-C coating:  $a - \times 10,000$ ;  $b - \times 100,000$ 

nanocrystalline tungsten carbide TiC takes 80 % in the coating and 20 % is the amorphous carbon matrix.

Amorphous carbon state in nc-TiC/a-C coating is assessed by Raman scattering spectra (Figure 7).



Figure 6. Spectrum of photoelectrons of nc-TiC/a-C coating (for designations see the text)

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In the spectra of all the samples two Raman scattering bands are observed, which are characteristic for the state of carbon with  $sp^2$ - and  $sp^3$ -



**Figure 7.** Spectra of Raman scattering on nc-TiC/a-C coatings of various composition: 1 - 8TiC; 2 - 19TiC; 3 - 26TiC



**Table 2.** Frequency positions ( $\omega$ ), pulse duration on half-amplitude level (FWHM), intensity ratio for D and G bands ( $I_D/I_G$ ) evaluated by RS spectra of studied structures (acc. Figure 7)

Sample (coating	D-band (sp <sup>2</sup> )		G-band (sp <sup>3</sup> )			I /I	
composition, at.%)	$\omega$ , cm <sup>-1</sup>	FWHM, cm <sup>-1</sup>	I, pulse	$\omega, \ cm^{-1}$	FWHM, cm-1	I, pulse	$I_D / I_G$
8TiC (42.5 C, 57.5 Ti)	1392.3	322.5	538.5	1577.7	113.1	403.4	1.33
19TiC (54 C, 46 Ti)	1397.0	346.7	396.9	1566.1	114.2	219.7	1.68
26TiC (70 C, 30 Ti)	1386.9	296.8	165.4	1565.4	132.6	134.5	1.23

Table 3. Mechanical properties of nc-TiC/a-C coatings ( $\delta$  = 3.2 µm) on substrates from 08Kh18N10T and Kh12M steels

Sample, substrate	Hardness <i>H</i> , GPa	Contact modulus of elasticity E <sup>*</sup> , GPa	Normalized hardness, <i>H / E</i> *, GPa	Modulus of elasticity <i>E</i> , GPa	$\begin{array}{c} \text{Stress of} \\ \text{out-of-contact} \\ \text{elastic strain } \sigma_{\text{es}}, \\ \text{GPa} \end{array}$	Coating composition, at.%	Grain size D, nm
29TiC, 08Kh18N10T	38.0	291	0.134	375	11.89	54 C, 46 Ti	2.9-4.3
30TiC, Kh12M	30.5	266	0.114	333	22.60	54 C, 46 Ti	2.9-4.3

hybridization of carbon electronic orbitals (bonds). G band of Raman spectrum with  $sp^3$ bond is due to the presence of an ordered graphite phase. D band of the spectrum with  $sp^2$  bond is characterized by structurally disordered graphite. Gauss approximating functions were used to perform decomposition of G and D band shapes into components, and ratio of band intensities  $I_D/I_G$  was determined. This ratio characterizes the proportion of graphite phases and is inversely proportional to the value of sp<sup>3</sup>-phase fraction in the coating [19]. Therefore, the most ordered amorphous carbon with greater content of sp<sup>3</sup>phase is found in nc-TiC/a-C coatings on 8TiC and 26TiC samples, in which the smallest values of  $I_D/I_G$  ratio are equal to 1.33 and 1.23 (Table 2).

Microindentation method [18] was used to measure the mechanical properties of coatings on substrates from 08Kh18N10T and Kh12M steels (Table 3).

High hardness of the coating (30.5–38.0 GPa), according to the data of other researchers [14], is related to its low content of the amorphous phase (a-C = 20 at.%). Normalized hardness values  $H/E^*$  allow evaluation of the level of coating resistance to plastic deformation, which rises with increase of  $H/E^*$ , which is an index of coating material wear resistance.

#### Conclusions

1. Method of simultaneous DC magnetron sputtering of graphite and titanium targets on to substrates from 08Kh18N10T, Kh12M steels and titanium VT1-0 was used to produce at deposition rate of  $1.4-2.2 \ \mu\text{m/h}$  a nanocomposite nc-TiC/a-C coating 2–3  $\mu$ m thick with TiC inclu-

sions of 3-10 nm size in an amorphous carbon matrix.

2. A procedure was developed for controlling nc-TiC/a-C coating composition by variation of power of magnetron discharge with a titanium target at constant power of the discharge with graphite target, thus ensuring the possibility of producing coatings in the composition range of 42.5-70 at.% C and 57.5-30 at.% Ti.

3. It is shown that the size of TiC grain and coating hardness depend on Ti/C ratio. Minimum size of TiC grain (2.9-4.3 nm) and maximum hardness (up to 30–38 GPa) are achieved at Ti/C ratio (in at.%) of 46/54.

4. Methods of Raman and photoelectronic spectroscopy were used to study the fine structure of nc-TiC/a-C coatings. It is found that TiC takes 80 % of the coating structure and 20 % is amorphous carbon matrix. It is shown that the degree of amorphous carbon ordering as to the content of  $sp^3$ -phase depends on coating composition.

5. Method of microindentation was used to determine the mechanical properties of nc-TiC/a-C coatings produced on various substrates. It is shown that the maximum normalized hardness of 0.134, which is a characteristic of coating material resistance to plastic deformation, was achieved on a substrate from 08Kh18N10T steel.

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### MELTING OF ELECTRODE AND BASE METAL IN ELECTROSLAG WELDING

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The present article is devoted to the problems of study of physical nature of electroslag welding using development of new methods and application of existing ones for investigations of processes of fusion welding. The results of study of electroslag welding process by wire electrode were given using method of direct visual observations through the optically transparent medium of phenomena running in the welding space and further shot-by-shot processing of materials of rapid filming. The analysis was given and description of some phenomena observed in the selected basic cell of welding space: melting of slag and electrode, formation of central nugget of an interelectrode gap in the form of slag-metal-gas plasma-type discharge, evolution of heat power and its dissemination in welding space and also numerous sizes of its basic geometric parameters. The obtained detailed conceptions about the physical nature of electroslag process allow more efficient using of its advantages in the development of new technologies and equipment for manufacture of thick-sheet massive welded metal structures. 28 Ref., 1 Table, 7 Figures.

**Keywords:** electroslag welding, welding space, slag pool, interelectrode gap, maximum temperature zone, active zone, interelectrode gap nugget, slag-metalgas plasma-type discharge, melting and transfer of molten metal, weld formation, rapid filming and photography

Since the invention and successful application of electroslag welding (ESW), being always the object of thorough attention among the scientists, more than fifty years have passed. At the initial stage the technical and economic advantages of this method predetermined often the priority of development and implementation of equipment and technological modifications of ESW.

However, the first investigations carried out at the dawn of development of ESW by the colleagues of the E.O. Paton Electric Welding Institute [1–18] and other research, industrial enterprises of the USSR [19-23], and also foreign specialists [24, 25] allowed the theoretical grounding of application of technological methods and temperature-time conditions as applied to different industrial structures. In the first turn the works on stabilization and control of welding process [6], automatic adjustment of level of metal pool [10, 12, 15] should be noted. The same important were the investigations of effect of welding conditions on quality of metal of welded joint [2], study of temperature field and thermal cycle and also heat balance of the welding process [14, 17, 19, 20, etc.].

Today ESW is challenging in the heavy machine building, especially in manufacture of welded metal structures of large thicknesses [25– 28], therefore the problems of improvement of efficiency of use of heat energy in melting of filler and base metal, optimization of methods of monitoring and control of ESW and also activation of investigations of this process remain still urgent.

The theoretical and practical conceptions about the phenomena occurring in ESW (running mostly at the closed isolated space), obtained earlier, were mostly based on the application of methods of indirect observation [2, 5, 6, 11]. To have a look inside the welding space (in direct meaning) became possible due to rapid filming and photography through the heat-resistant optically transparent medium [18], which is installed instead of copper forming device, cooled by water (Figure 1). As the welding current, passing through the slag pool, makes it opaque, to see the real character of electrode melting and transfer of molten metal into the metal pool became possible only at maximum approach of wire electrode directly to the surface of a quartz glass. Thus, through the surface of transparent medium the plane image (projection of cross section) of welding space is seen, which is located in the plane of moving electrode approached to the glass, i.e. the processes of melting and transfer of electrode metal in the slag pool are observed.

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One can assume that the processes like these will run also in the melting zone of the second electrode closed from observation of slag pool. The further investigations of EW process using wire electrode were carried out basing on this procedure, including:

• direct rapid filming and photography of ESW process through the optic transparent medium of specimens of low-alloyed steel of 09G2S grade 60 mm thick using two wire electrodes of 3 mm diameter at the 45 mm slag pool depth, welding gap equal to 27 mm, dry 70 mm electrode stickout, under the flux AN-8 (see Figure 1). The supply was made from AC source of TShS-3000/3 type with a rigid external characteristic. The welding process was running at the rates of wire electrode feed equal, respectively, to 3.11 and 4.39 cm/s;

• study of dynamics of change of geometric sizes of basic parameters of welding space using computer processing of frames of filming the electroslag process;

• analysis of basic electric parameters of welding process  $(I_{\rm w}, U_{\rm w}, v_{{\rm w.f}})$  and temperature mode in welding space. For this purpose, at keeping all the conditions of ESW fulfillment, recorded earlier in filming and photography, the welding of specimen with computer high-frequency record of  $I_{\rm w}$ ,  $U_{\rm w}$ ,  $v_{{\rm w.f}}$  and measurement of temperature of slag and metal pool was performed;

• the comparison of data on geometric parameters of welding space with the basic electric parameters of welding process and also temperature of slag and metal pool for establishment of frequency-temporary coincidences of a pulsed nature of electrode metal transfer and changes of welding current.

Below the results of visual investigation of welding space in ESW using wire electrode (first and second items) in isolated closed zone formed by edges being welded, forming devices, weld and mirror of slag pool, where melting of electrode and base metal as well as weld formation occur, are given.

For the selected welding conditions one wire electrode is intended for 30 mm thickness of metal being welded. Therefore as an optimal cell per one electrode of 3 mm diameter the welding space was selected limited by the sizes  $B \times S \times h_{\rm sl}$ , where B = 27 mm – width of welding gap; S = 30 mm – thickness of metal being welded;  $h_{\rm sl} = 45$  mm – depth of slag pool.

The analysis of visual observations of the welding space (Figure 2) confirms that in this most responsible link of electric circuit, representing the concentrated ohmic resistance, the main heat energy is formed and evolved which is further transferred to the electrode and base metal.

In the welding space, the longitudinal section of which in the plane of electrode axis conditionally resembles the shape of turned-over «mushroom», the single, usually visually non-visible components (see Figures 1 and 2) — slag pool, metal pool, line of solidification front, weld and wire electrode — can be distinguished.

In general, all the parameters of welding space, indicated in Figure 2, are correlated and continuously changed with time. It is clearly seen that between two solid metallic conductors (base metal and wire electrode) the electric conductor in liquid form constantly exists. In the first approach it should be noted that its central part can be defined as a nugget of the interelectrode gap. The sizes and shape of a weld are mainly determined by the amount and character of heat dissemination in welding space, here, the edges of base metal are fused higher than the level of metal pool mirror.

The slag pool, representing the melt of mixture of oxides, salts, sulfides and other components, is a conductor of electric current and is governed by the Ohm law. It is known that in the slags the ion conductivity prevails [1, 2, 10]. In the total volume of slag pool, having, as compared to the metal being welded, the considerably higher Ohm resistance in general and in the central nugget of interelectrode gap, in particular, the conversion of electric power into the heat one is occurred.

The upper boundaries of slag pool are distinctly outlined by the mirror of its almost flat surface. In the gap formed by the base metal edges being welded, such separation between slag and metallic pool is not observed (Figure 2). The slag pool in its volume is mainly, as was expected, rather non-uniform in its temperature, which is proved by its color gamma (see Figure 2). The typical regions can be distinguished in it:

• the region of the highest temperatures (area  $F_1$ ), which directly contacts the edge of electrode, where the overheat of drops of molten electrode metal and slag is occurred;

• more branchy region (area  $F_2$ ), which is characterized by lower temperature than area  $F_1$ , but higher than the temperature of basic volume of slag pool. Let us define this region as the active region, where heating and melting of electrode occur.

Both these zones in welding space occupy the volumes which are definite and, obviously, optimal, from the point of view of heat energy evo-





**Figure 1.** Scheme of specimen for rapid filming and photography of ESW through optically transparent medium: l – edges being welded; 2 – wire electrode; 3 – nozzle; 4 – forming device (shoe); 5 – slag pool; 6 – metal pool; 7 – weld; 8 – quartz glass (instead of reverse shoe);  $L_{\text{wet}}$  – electrode wet stickout;  $L_{\text{w.p}}$  – length of wetted electrode part

lution and stability of welding process. The observed zones  $F_1$  and  $F_2$  should be considered and represented as longitudinal sections of the corresponding volumes  $V_1$  and  $V_2$  close to the bodies of rotation of mentioned plane sections around the axis of electrode.

The volume of zone of the highest temperatures  $V_1$  is actually a central nugget of the interelectrode gap. It is obvious that its composition, shape and state are the most important characteristics of the electroslag process. At the selected parameters of welding mode the nugget (see Figure 2) exists during the whole period of welding changing in volume in a pulsed form according to the definite laws. The visual observations of the process give grounds to assume that according to its composition and state the volume of a nugget  $V_1$  represents some slag-metal-gas discharge of plasma type, which is formed in slag pool as a result of current passing through it. It is formed and exists in a pulsed mode. It is clearly seen that this discharge has a considerably higher temperature than the rest regions of slag pool, and as a conductor of electric current is subjected to the influence of electromagnetic fields generating in the welding circuit. The physical state of this zone (e.g. temperature, conductivity, etc.) should be determined using special physical methods of investigation, which are applied in study of discharges of the kind.



**Figure 2.** Scheme of distinguish of zones of space of formation of welded joint of ESW using wire electrode fixed at rapid filming through optically transparent medium and their designation: B — width of gap between the edges being welded; e — weld width; l — distance between the electrode end and metal pool mirror;  $h_{\rm m}$  — metal pool depth;  $F_1(V_1)$  — zone of the highest temperatures;  $F_2(V_2)$  — active zone;  $B_1$  — width of zone of the highest temperatures;  $B_2$  — width of active zone

In Figure 2 one can see how alternating welding current (for example, in the first semi-period) is transferred from the electrode to a slag pool. The contacting metallic hard surface of electrode can considerably change in this case: from the size equal to cross section of edge of electrode  $(7.1 \text{ mm}^2)$ , to the value of general area of side surface of wet electrode stickout (about  $100 \text{ mm}^2$ ) which contacts (wetted) and fused by a slag. Therefore, on the contact boundaries of volumes  $V_1$  and  $V_2$  the density of current and conductivity are changed continuously. Welding current is spreading in the volumes of these zones passing through the contact surface with mirror of metal pool and partially through the fused edges being welded, arranged over the metal pool. From the metal pool through the surface of solidification front and fused edges of base metal the electric circuit is closed to the base metal and then further to power source. Here, the area of contact surface of base metal is equal to the area, which is wetted by the molten metal and slag. It many times exceeds the contact surface of wire electrode. Therefore during change

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**Figure 3.** Constantly existing current-conducting channel (boundaries of channel are marked with arrows) from zone  $F_1$  to metal pool, the sizes of which are determined by its maximal width  $B_1$  (see Figure 2)

of polarity (second semi-period) the conditions of welding current passing are changed. This phenomenon is visually observed by alternation of change of brightness of glowing of adjacent frames. Here, around the perimeter of zone of the highest temperatures  $V_1$ , coaxially to the electrode, the bright current-conducting channel, contacting the metal pool, is constantly observed (Figure 3).

During passing of welding current the basic part of heat energy is mostly evolved on the contact boundaries (see Figure 2): surface of electrode-region of volume  $V_1$ ; surface of electroderegion of volume  $V_2$ ; region of volume  $V_1$ -metal pool; region of volume  $V_2$ -metal pool; metal pool-base metal (edges + weld); slag pool-edges of base metal.

It is obvious that at the constant existence of several contact zones between liquid conductors with different properties (for example, conductivity, temperature, viscosity) in the interelectrode gap, some concentration of heat energy, evolved in the central nugget of interelectrode gap, is occurred.

Moreover, the vector of movement of heat energy of region of volume  $V_1$  is mostly directed towards the metal pool, as its heat conductivity



**Figure 4.** Schemes of distinguishing the zones  $F_1(a)$  and  $F_2(b)$  at the separate frames of filming at calculation of their areas and volumes



is higher than that of a slag. The slag pool and edges being welded receive the main heat pulse from metal pool as the most moving medium, formed in welding space. The heat energy transferred to slag pool is consumed for preheating and partial fusion of edges being welded, wet electrode stickout and also heating of slag pool. In addition, a part of volume of slag pool, located higher than active zone  $V_2$ , is not a conductor of welding current, i.e. it does not mainly evolve, but consumes heat energy and provides equilibrium state of  $F_1$  zone and also protection of metal pool from the atmosphere effect.

In determination of fraction of areas (volumes) of zones  $F_1$  and  $F_2$  in total balance of selected cell one can divide them conditionally for convenience into more simple plane geometric figures (Figure 4) and calculate their areas. As it was mentioned earlier, they are actually the cross sections of bodies (volumes) of rotation relatively to the axis of electrode, their areas can be converted into the proper volumes. Examples of changes of these areas with time, as well as correlation of volume  $V_1$  to the total volume of selected cell of welding space V, are described in Figure 5.

For the selected time interval, for example, of 0.45 s, the area of zone  $F_1$  (Figure 5, *a*) can change within wide ranges (from 25 to 200 mm<sup>2</sup>), the area of  $F_2$  zone changed at the same time interval within the ranges of 350–550 mm<sup>2</sup> (Figure 5, *b*). The fraction of volume of nugget of interelectrode gap  $V_1$  to the total volume of selected cell of welding space V is changed negligibly (Figure 5, *c*, *d*). These data prove that main heat energy, evolved in ESW, is really concentrated in the volume of zone of the highest temperatures  $V_1$ .

The metal pool is formed of molten electrode metal, which is supplied into it mostly in portions from the volume  $V_1$  (zone  $F_1$ ) and base metal fused along the edges. It is established that at stable proceeding of the electroslag process the alternation of relatively calm character of formation and movement of mass of volume  $V_1$  into metal pool, and its explosive-like transfer is observed. The explosive type character of transfer of electrode metal is a consequence of some pulsed accumulation of heat energy in the overheated nugget with the subsequent explosive type discharge of the inerelectrode gap. As a result, at the selected modes of ESW the local advance accumulation of evolved heat energy and comparatively its delayed consumption (heat removal) by the edges of base metal is observed. This phenomenon influences greatly the shape of interelectorde gap and character of drop transfer.



**Figure 5.** Changes of areas  $F_1(a)$ ,  $F_2(b)$  and fraction of volume of zones of the highest temperatures  $V_1$  in total volume of welding zone at  $v_{w.f} = 3.11$  (c) and 4.39 (d) cm/s

The accumulated heat leads to increase of rate of electrode melting, causing pulsed growth of conductivity and mass of volume  $V_1$  and also the welding current. When under the influence of heat and electric factors the powerful explosive discharge of  $V_1$  volume occurs, the value of wet stickout decreases (parameter l (see Figure 2) increases) and, as a consequence, the value of welding current is decreased. After that the general cycle is repeated and origination, formation and growth of a new volume  $V_1$  begins. The frequency of formation and volume of slag-metal-gas discharge of plasma type depends on selected parameters of welding conditions. Thus, at the rate of electrode feed of 4.39 cm/s during the preset welding period of duration of one discharge equal to 0.15 s, 0.24 g of electrode mass was melted, and speed of dissipation of heat energy (wave) in metal pool, i.e. the movement of heat flow of volume  $V_1$  mass, amounted to about 1.5 m/s.

If the metal pool has a clear, slowly changed interface with weld (along the line of solidification front), then such stable interface between

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**Figure 6.** Dynamics of change of metal pool shape at  $v_{w,f} = 3.11 \text{ cm/s}$  at the beginning (a, b) and steady process  $(c, d): 1 - \text{zone } F_1$  formation; 2 - explosive-like penetration of mass of volume  $V_1$  into the metal pool



**Figure 7.** Change of shape of metal pool at  $v_{w,f} = 4.39 \text{ cm/s}$  at the moment (dotted lines mark the boundaries of metal pool) of origination of volume  $V_1$  (*a*), beginning of «forcing» of mass of volume  $V_1$  into the metal pool (*b*), growth of volume  $V_1$  mass (*c*), and explosion of mass of volume  $V_1$  with forcing out of metal pool (*d*)

the slag pool and mirror of metal pool in the process of melting is too difficult to determine. During the whole electroslag process the mirror of metal pool (especially under electrode) has a complicated concave conical surface which changes constantly (Figures 6 and 7). Under the influence of mass of volume  $V_1$  and occurring hydro- and electrodynamic forces, the shape of surface of metal pool mirror changes constantly. At the moments, when the mass of volume  $V_1$ reaches bottom of metal pool (surface of front of weld solidification) by the pulsed energy of powerful directed discharge, one can observe the «splashing» of metal pool into the slag beyond the limits of equilibrium state (Figure 7, d). As a consequence, the overheated slag-metal-gas mixture together with metal pool moves upwards along the planes formed by the front of solidification and forming devices. Then, then the molten metal is flowing down already under the effect of gravity forces. The intensive transfer of heat energy to the slag pool and edges being welded occurs by alternating pulses. The heat pulse from metal pool is directed to the earlier fused edges of base metal under the angle of about 90°, therefore it transfers them maximum energy. The upper part of edges in the gap, contacting the slag pool and being above the level of end of electrode, is heated less intensively and, as a rule, does not fused by a slag. As even skull crust can exist at the boundary, then mainly due to this the mushroom shape of edges penetration is observed.

The line of solidification front practically repeats the shape of mirror of metal pool. The concept of depth of metal pool, usually determined from the macrosections, does not meet its real parameters during running of electroslag process (see Figures 6 and 7). It should be noted that periodic «splashing» of portions of metal pool causes partial repeated fusion of crystallized weld metal along the solidification front.

The intensive rotation of mass of volume  $V_1$  around the axis coaxial to the electrode is also observed. As a result the heat flows can deviate aside from the axis of electrode. However the main fraction of heat energy is concentrated along the axis of electrode movement, therefore the surface of crystallization front is maximum concave along the weld center in particular.

The melting of electrode metal occurs mainly in the region of active zone  $F_2$ . Moreover, the electrode is fused along its side surface, wetted by the slag (contacting the active zone). The fused cone-shaped surface of end of metal electrode is the main contact element of electric cir-



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cuit, through which the welding current is directed to slag pool from the electrode.

The drop formed from the molten metal, which is flowing along the side surface of electrode, is directly in the zone of highest temperatures  $F_1$ , where its additional overheating occurs. During formation of drop it is affected by the gravity force and electrodynamic forces («pinch-effect», accelerating the moment of detachment and movement of drop from electrode).

It was established that the shape of end of the electrode during the welding process is mainly cone-shaped (the electrode is sharpened downwards). However at the moment of powerful pulse explosion of discharge sometimes lump-like rupture of end of conic part of electrode is observed, as a result of which the wet stickout is noticeably decreased, and the distance l is increased. It is rationally to add that in the so-called wet stickout  $L_{wet}$ , which defines the depth of submersion of electrode into the slag pool, the part should be specially distinguished wetted by the slag  $L_{w.p}$  (see Figure 2), influencing the welding current and character of electrode melting.

For the investigated conditions the alternation of three-four and more small discharges with one-two big explosions was noted. The shape of edges fusion is explained by the character of heat transfer from the metal pool to the base metal and partially by the conditions of existing of metal pool under the effect of gravity force and forces of surface tension. The contour of edges penetration represents the isothermal surface corresponding to the melting temperature of base metal. In the zone of electrode metal melting the intensive gases evolution is observed, which are evolved practically through the whole volume to the mirror of slag pool (see Figure 3). The evolution of gases is observed also in the region of maximal penetration of edges of base metal (see Figure 3, frame 6378), which can be explained by getting of gases to the metal pool with the mass of volume  $V_1$  (at the moment of explosion). The evolution of gases at the electrode was not visually noted.

In study of welding zone the different computer programs of precise discrete measuring were used, by means of which the values of geometric parameters were determined (see Figure 2). The sizes were determined during processing of fragments of filming frames, performed at the frequency of 100 frames per second for two modes of ESW – at  $v_{w,f} = 3.11$  and 4.39 cm/s, respectively. The determination of real linear sizes of geometric parameters was performed as applied to known set value of real welding gap

 $v_{\rm wf}$ , cm/s Parameter (see Figure 2) and dimensions 3.11 4.39h<sub>m</sub>, mm 4 - 104 - 16 $l + h_{\rm m}$ , mm 20 - 22.520 - 26.830.5-33.5 30.5-40.5 L, mm48-51.5 54-60 e, mm  $F_1$ , mm<sup>2</sup> 100-210 20-200  $F_2$ , mm<sup>2</sup> 190-550 \_

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Real limits of change of numerical dimensions of main geometric

parameters for the selected cell of welding space depending on

wire electrode feed rate

 $B_1$ , mm

 $B_2$ , mm

between the edges B = 27 mm. The error of measurements amounted to 0.5–2.5 %. For clearness some results of measurements (the limits of parameters values) are given in the Table.

It is known that the main criterion for evaluation of quality of a weld is the coefficient of weld shape  $\psi = e/h_{\rm m}$ . Though the depth of metal pool in the process of welding is changed continuously, the width of weld *e* is changed negligibly. The section area of metal pool can amount to 30 % of area of plane of section of welding space (along the axis of electrode) and its mass for the mentioned cell is in the limits of 115– 125 g. It is the main factor of influence on the weld shape. For investigated modes of welding the average value is  $\psi = 5.7-7.0$ .

#### Conclusions

1. At the stable process of ESW the wire electrode is fused in a slag pool along the surface wetted by a slag. During getting of molten electrode metal into the gap between the electrode and mirror of metal pool, the slag-metal-gas plasmatype discharge is formed, which is transferred to metal pool by alternating pulses of overheated mass by electro- and hydrodynamic impacts. Reflecting from the bottom, the overheated mixture, including metal pool, is lifted to the edges being welded and transfers heat energy to them and also to a slag pool.

The explosion character of transfer and intensive mobility of overheated slag-metal-gas mixture provide a mushroom shape of penetration of edges being welded.

2. The presence of cyclic repeated slag-metalgas discharges of plasma type, observed during investigations, is the character feature of electroslag process using wire electrode at all the stages of its existence.



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3. Slag pool in ESW is rather non-uniform in temperature, which is evidenced by its fixed color gamma, where characteristic regions can be distinguished:

 slag-metal-gas discharge of plasma type (volume  $V_1$ ), which contacts directly the electrode and has the highest temperature;

• active zone (volume  $V_2$ ), which as compared to  $V_1$  is characterized by lower temperature, but being higher than the temperature of main volume of slag pool;

• in these regions the electric energy is transformed into a heat one, the melting and transfer of electrode metal, as well as weld formation, are occurred;

• part of volume of slag pool, located over the zone  $V_2$ , provides equilibrium state of region  $V_1$ , protection of metal pool from the effect of atmosphere and is not a conductor of welding current in general.

4. The considered regions of the welding zone are the most important elements of electroslag process and together with electric parameters of welding conditions are used in monitoring and control of welding process.

5. Peculiarities of running of ESW using wire electrode can be taken into account during investigation of such welding methods as welding with consumable nozzle, with large-section electrode, and also electroslag remetling.

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# ACQUISITION OF PROCESS IRREGULARITIES BY MEANS OF ACOUSTIC DISTORTION PARAMETERS DURING GMA WELDING PROCESSES

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In GMAW, process irregularities often result in weld defects and have, thus, negative effect on the weld quality. Currently, the online detection of these irregularities in industrial production is mainly based on the analysis of current path and voltage time curves. Arc length control, which, for example, is used in modern welding machines, evaluats the relative changes of electric process parameters (welding current and voltage) for detection of variations of the arc length. Superimposing interferences are, however, complicating the unambiguous acquisition of arc length via electric parameters. Within the scope of research project, which has been carried out in the Welding and Joining Institute, this problem has been solved by establishing the unambiguous link between length of the arc and acoustic distortion parameters, which are the result of using an arc as a sound converter. For the detection of process irregularities, defined modulation of welding current has been carried out, the changes of which during welding have been acquired and analysed. Similar as in real audio loudspeaker, distortions during reproduction are developing, which are acquired by a directional microphone and subsequently evaluated. The parameters THD (Total Harmony Distortion) and SINAD (Signal-to-Interference Ratio including Noise and Distortion) have turned out to be the most reliable electro-acoustic parameters. Changes of arc geometry exert unambiguous and reproducible effects on these parameters. Vice versa, it is proceeded from the assumption that, in the case of persistence of parameters within a certain corridor, the arc has not been deformed. For the acquisition of arc length, the modulation of arc and acoustic acquisition and evaluation offer thus an unambiguous alternative and/or completion to the method, which has been used so far and is based on the electric parameters. The measuring set-up and test results of this novel measuring method will be specified in detail in this paper. 10 Ref., 6 Figures.

#### **Keywords:** GMA welding, quality control, acoustic emission, arc length control, audisignal distortion parameters, power sources

In GMAW, process irregularities often result in weld defects and have negative impact on the weld quality. The online detection of these irregularities in industrial production is, nowadays, mainly based on the analysis of current path and voltage time curves. Arc length control, which is used in modern welding equipment, evaluates, for example, the relative changes of electric process parameters (welding current and voltage) with regard to a target value for the acquisition of arc length variations. Superimposing disturbances impair, however, the unambiguous acquisition of the arc length via the electric parameters. Functionality of the arc sensor system is, moreover, based on the analysis of these process parameters [1-4].

Since, however, the quality of welding result is directly or indirectly dependent on temporal behaviour of the arc, determination of the effective arc properties during GMAW is of utmost importance [5, 6]. The weld geometry is, thus, with high degree influenced by the type of material transfer, heat input, arc pressure, which is exerted on the molten pool, and flow conditions in the molten pool, which are resulting thereof. Moreover, type of material transfer and resulting droplet properties, as well as burn-off of alloying elements from the droplet, is determined by plasma-physical conditions in the anode region of wire electrode, plasma column temperature and plasma composition and/or by the distance covered within the arc column. Welding defects such as undercuts, lack of sidewall fusion and lack of inter-run fusion are, moreover, caused by unfavourable arc attachment on the workpiece [7].

The model approach which has been used so far, i.e. the assumed proportionality of one-dimensional geometrical arc length to arc voltage, is insufficient, as depicted in Figure 1. At almost the same arc length a strongly different voltage is setting. In the present case, the 4 % reduction of the one-dimensional geometrical arc length is accompanied by 6 % increase of the voltage, that is clearly contrary to the currently applied model idea of proportionality of voltage and length of

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**Figure 1.** Non-proportionality of geometrical arc length and welding voltage

the arc. Against this background it is essential to add another fundamental process parameter, which is capable to provide further information on length and shaping of the arc.

Acoustic measurements of the arc. Ever since the «singing arc lamp» has been discovered at the Erlangen University in 1897 it has been documented that the arc is, basically, usable as a loudspeaker. It is capable to transmit frequencies within the range of 16 Hz up to range, which is no longer audible for human beings. This is basically possible in a very high quality, that is proven by the fact that in very pricy audio loudspeakers, which are on the market, the arc is used as sound converter for the tweeter range [8].

Applied to the welding technique it is, therefore, obvious to couple low-frequency audiosignals into welding power source and to evaluate the resulting sound signals. It was, therefore, required to test the hypothesis whether the geometrical shape of the arc exerts direct influence on acoustic parameters, which are capable to quantify the «transmitter system welding process». It stands to reason that, here in particularly, the acoustic parameters are to be considered, specified the non-linear distortion of the signal, to which the signal is subjected on the path from sound generation to the arc.

Figure 2 shows overall electric system, which consists of contact tube, wire, arc and workpiece. It is obvious that each subsidiary system exerts great influence on the electric behaviour. Analysis of the individual influences on resulting electric temporal behaviour is extremely complex. However, proceeding from the assumption that evaporation processes at anode, cathode and droplet, and also that the developing plasma and shielding gas flow dynamics do not cause noticeable acoustic emissions, it is only the arc system, which can be assumed to be a sound transducing subsidiary system.

In preliminary tests with TIG welding equipment, the proof has been delivered that the welding machine is, basically, capable to clearly reproduce coupled test sounds despite clearly audible process noise. Figure 3 shows spectrum of triad of three equally loud sounds of the 2, 5 and 7 kHz frequencies which have been modulated on an inverter power source. Besides the process and background noise, the low-pass behaviour of used transmission system is clearly recognizable.

In preliminary tests, the THD (Total Harmony Distortion) and SINAD (Signal-to-Interference Ratio including Noise and Distortion) parameters have proven to be the most reliable electroacoustic parameters. Herewith, changes of the arc geometry had unambiguous and reproducible effects.

THD parameter is supposed to be an important quality criterion in the generation and transmission of signals. Just as every quadripole,



Figure 2. Operative path of arc welding process with idealised potential curves



Figure 3. Spectrum of triad coupled into inverter power source



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which is made of non-linear components, welding process also represents a non-linear system. If a sinusoidal signal is applied to the non-ideal characteristics of the process, inevitably harmonics are developing, which are found at the integral multiple frequencies of the base frequency (harmonic). The THD guantifies the frequency approximately a

acteristics of the process, inevitably harmonics are developing, which are found at the integral multiple frequencies of the base frequency (harmonic). The THD quantifies the frequency components of this harmonic, which is, as the rule, unwanted. For the calculation, a relation is established between the sum of effective value of harmonics and the sum of effective values of fundamental wave plus harmonics:

THD = 
$$\left(\frac{\overline{U_2^2 + U_3^2 + \dots}}{\overline{U_1^2 + U_2^2 + U_3^2 + \dots}}\right)^{1/2}$$
.

The THD is dimensionless, and is expressed alternatively in % or, logarithmical, in dB [9].

SINAD parameter is the ratio of root-meansquare signal amplitude to mean value of the root-sum-square of all other spectral components, including harmonics, but excluding DC-parts of the signal. SINAD is the good indication of the overall dynamic performance of system because it includes all components, which make up noise and distortion. Assuming signal S, noise N and distortion D are measured with the same input signal amplitude and frequency, SINAD is calculated as follows and is expressed in dB [10]:

$$SINAD = 20 \log\left(\frac{S}{N+D}\right).$$

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**In-situ-measurement during welding.** For the test set-up, welding power source Cloos Quinto II with wire feeder has been used. Since welding power sources are switched, the external supply of sinusoidal signal is impeded considerably. In order to circumvent this problem in the test set-up, welding current supply signal of the Quinto II has been tapped and coupled with a function generator via an summer circuit. The resulting mixed signal is, subsequently, amplified with longitudinally controlled transistor power source ELMA 800, which is, in this case, used as fast and inverter-ripple free power section and supplied to the welding process (Figure 4).

Deposition tests using the spray arc with the contact–tube–distance of 18 mm have been carried out. Shielding gas of M21 class in accordance with DIN EN ISO 14175 with 18 %  $CO_2$  was used.

After short settling time, quasi-stationary welding process was set. Now, sinusoidal signal, which has been generated by the function generator, was coupled into the process. The sound, which, thereupon, emitted by the spray arc, was acquired via directional microphone with a distance of 1 m, digitalised and analysed online with PC.

The tests had, in the first step, been carried out in the audible range in order to guarantee the immediate acoustic control of coupled sig-



Figure 4. Test set-up of welding equipment



**Figure 5.** Chronological sequence of SINAD (*a*) and THD at welding voltage of 26, 29 and 32 V

nals. Tests made with frequency analysers showed, however, clearly that measurements also in the range, which is no longer perceivable for human beings (18–20 kHz) will be possible in the future.

Since length and shape of the arc column are in direct connection with the wire feed rate and welding voltage, a series of measurements with variations of voltage in the range 20-40 V have been carried out in order to achieve different geometrical shapes of the arc column at the constant wire feed rate.

**Results.** The variation of welding voltage has immediate effects on the arc, namely with increasing voltage the arc length is also increasing and extends thus the surface, that is effective for sound generation.

The tests have established that the SINAD and THD parameters are decreasing with longer arc length. Figure 5, *a* shows, by way of example, chronological sequence of SINAD depending on welding voltage. Here, wire feed rate had been constantly 10 m/min, welding current was modulated with the 12 kHz sinusoidal signal. Figure 5, *b* depicts chronological sequence of THD at various welding voltage and constant wire feed rate of 10 m/min. The frequency of the coupled sinusoidal signal was also within the range of 12 kHz.

Figure 6 shows variation of the arc length via the change of wire feed rate. Welding voltage was constantly 30 V, coupled sinusoidal oscillation had the frequency of 12 kHz. In line with the Figures above, here also the trend of decreasing distortion parameters (THD and SINAD) with increasing arc length is observed.

**Summary and perspectives.** The use of arc as a sound converter and the measurement of welding process distortion parameters, became useful



Figure 7. THD and SINAD versus wire feed rate



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addition to existing characterisation of arc with regard to the arc length, which is based on analyses of the transient current and voltage signal. The most reliable parameters have proven to be the THD and SINAD, and changes of the arc geometry take unambiguous and reproducible effects.

In-situ measurement of the process with comparatively low technical expenditure is possible. Just the coupling of external signals is problematic in welding power sources due to their low inverter pulse frequency. The next generation of welding power sources will, however, be most likely equipped with increased inverter pulse frequencies since this is required for adhering to the Nyquist Shannon sampling theorem.

The tests have been carried out in the form of metal deposition using the spray arc. Further tests are planned with the inclusion of all other welding processes in order to verify the relevancy to practice and the robustness of method.

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### BRAZING FILLER METALS OF Ti-Zr-(Fe, Mn, Co) SYSTEM FOR BRAZING OF TITANIUM ALLOYS

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Titanium alloys are perspective materials for different branches of industry. Appearance of new high-strength materials, in particular intermetallic alloys, provides for increasing interest to processes of their joining by brazing methods. Meanwhile, the most wide spread brazing filler metals (Ti–Cu–Ni and Ti–Zr–Cu–Ni systems) developed decades ago do not always correspond to current requirements, as, for example, in brazing of intermetallic alloys. Present work provides the results of complex investigations of brazing filler metals of Ti–Zr–Fe, Ti–Zr–Mn and Ti–Zr–Co systems using differential thermal analysis, light and scanning microscopy, X-ray microspectrum analysis. Data on melting ranges of pilot alloys were obtained, and liquidus surfaces of given systems using simplex-lattice method were build. Brazing filler metals covering brazing temperature range of current structural titanium materials based on solid solutions as well as intermetallics were proposed. Structure, chemical inhomogeniety and strength characteristics of brazed joints were studied. It is determined that brazing of solid solution based alloys (OT4, VT6) using indicated brazing filler metals ensures strength characteristics of joints, which are not inferior to that obtained with application of known brazing filler metals. Proposed brazing filler metals provide strength on the level of base metal at room and elevated temperature as well as in creep-rupture testing during brazing of  $\gamma$ -TiAl intermetallic based alloy. 13 Ref., 4 Tables, 8 Figures.

**Keywords:** vacuum brazing, titanium alloys, intermetallic alloys, brazing filler metals, brazed joints, structure, strength of brazed joints

Area of application of welded structures from titanium and its alloys constantly expands with rise of volume of its manufacture and cost reduction. It is of course promoted by favorable combination of mechanical and special properties of titanium, among which, first of all, are its low specific weight, high strength and corrosion resistance. Undoubtedly, welding takes the leading role in development of titanium structures. However, in number of cases technological processes of brazing are more appropriate and, sometimes, being the single possible, in particular, during production of multilayer thin-wall structures. Appearance of new intermetallics based highstrength titanium alloys also increases probability of application of brazing technology. This explains constant attention of high range of specialists to development of brazing filler metals (BFMs) for brazing of titanium alloys and methods of their manufacture in appropriate for application form.

It should be noted that BFMs of Ti-Cu-Ni, Ti-Zr-Cu-Ni, Zr-Ti-Ni and Cu-Zr-Ti systems in a form of plastic foils, obtained by method of ultraspeed quenching or traditional methods of metallurgical redistribution with pressure (rolling) treatment and vapor phase deposition, as well as in powder form [1–5] are mainly used in world practice for brazing of titanium alloys.

However, development of new alloy systems is continued. It is related with the tasks of reduction of brazing temperature for wrought titanium alloys, as well as expansion of area of BFM application (for example, in medicine, in brazing of intermetallic alloys etc.). It should be noted that reduction of temperature of brazing of wrought titanium alloys by BFMs of existing systems decreases, as a rule, strength characteristics of the brazed joints.

Present work proposes the braze compositions selected on the basis of complex investigation of Ti–Zr–(Fe, Mn, Co) system alloys, which can be used for brazing of wrought and intermetallic titanium alloys providing temperature-time parameters of technological process of vacuum brazing and preserving microstructure and mechanical properties of initial material to be brazed, as well as eliminating formation of brittle intemetallic phases in metal of the seams.

Studies of Ti–Zr–Fe, Ti–Zr–Mn, Ti–Zr–Co system alloys [6–8] were carried out at the E.O. Paton Electric Welding Institute as an alternative to existing ones. Constitutional diagrams of Ti–Fe, Ti–Mn and Ti–Co systems are similar. Eutectics with high content of titanium and wide area of solid solution based on titanium and eutectoid are present in high-titanium area of these

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Temperature, °C	Ti-Mn	Zr-Mn	Ti–Fe	Zr–Fe	Ti-Co	Zr–Co
Eutectic melting	1180	1090	1085	928	1020	981
Eutectoid transformation	550	790	595	730	685	834

Table 1. Temperature of eutectic melting and eutectoid transformation of alloys of the various systems [9]

alloys. Ti-Mn system has the highest temperature of eutectic melting, Ti-Fe system displays significantly lower and Ti-Co has the smallest one (Table 1).

Typical characteristics described above are preserved in the alloys of Zr-Fe, Zr-Mn and Zr-Co binary systems [9]. At the same time, areas of solid solutions are narrower, and eutectoid transformation takes place at higher temperatures. Melting temperatures of eutectics follow the tendency of indicated titanium-based alloys except for Zr-Co system (see Table 1).

It can be assumed based on study of binary systems that ternary eutectics with temperature suitable for brazing of titanium wrought pseudo $\alpha$  and  $(\alpha + \beta)$ -alloys (not more than 935 °C) and intermetallic alloys (above 1150 °C) are present in three-component systems. It was necessary to build liquidus surfaces of these three-component systems in order to verify this hypothesis. Combination of calculation and experimental methods, in particular, method of simplex-lattice planning of experiment [10, 11], was used for realization of this task. This method is developed for reduction of number of physical experiments, decrease of time consumption as well as costs. Field of application of present method is sufficiently wide and can be used for building of «composition-property» diagrams, liquidus surfaces





**Figure 2.** Microstructure (×500, light microscope) of central area of seam of brazed joint (OT4 base metal, Ti-Zr–Co BFM)

and surfaces of phase transformations in multicomponent systems etc.

From 33 to 57 alloys of each system were manufactured and their melting ranges were determined for obtaining of necessary calculation data. Results of calculations in graphical form are represented in Figure 1.

Analysis of obtained results shows that specific part of each from the three alloy systems, containing monovariant eutectics with decreased melting temperature, is the most suitable for application as BFMs in brazing of titanium and its alloys. As expected alloys of Ti–Zr–Mn system were the most refractory, and Ti-Zr–Co system had the lowest melting temperature. One BFM from each system was selected for investigation of their technological properties and strength of

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brazed joints in order to compare with known BFMs. BFMs of Ti–Zr–Fe and Ti–Zr–Co systems were used for brazing of structural titanium alloys (together with industrial BFM of Ti–Zr–Cu–Ni system) and Ti–Zr–Mn and Ti–Zr–Fe systems — for alloys on the basis of  $\gamma$ -TiAl compound.

Brazing of specimens was performed in vacuum (7·10<sup>-3</sup> Pa) with the help of radiation heating. Temperature of brazing of wrought titanium alloys OT4 (Ti–4Al–1Mn) and VT6 (Ti–6Al– 4V) using BFMs of Ti–Zr–Co and Ti-Zr-Fe systems equaled 920 and 990 °C, respectively. Brazing time made 15 min. Intermetallic titanium alloy (Ti–45Al–2Nb–2Mn + 0.8 vol.% TiB<sub>2</sub>) was brazed at temperature close to heat treatment temperature of 1250 °C with 60 min holding.

Results of performed experiments determined that BFMs in cast form spread well over the surface of titanium alloys and form smooth full fillets.

Metallographic investigations of the brazed specimens verify that external welds, which were brazed using selected industrial and experimental BFMs, have no significant differences. Seam in some distance from the fillet represents itself common intergrown grains of the base metal. Sometimes these areas are impossible to be distinguished from the base metal and joint zone can be determined only through investigation of chemical inhomogeniety (Figure 2).

Distribution of elements in the seam metal reflects significant leveling of concentrations



**Figure 3.** Microstructure of seam metal (*a*), and quantitative distribution of titanium (*b*), zirconium (*c*) and cobalt (*d*) over the width of brazed joint along scanning line (OT4 base metal, Ti–Zr–Co BFM)



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Ti-Zr-Co BFM), wt.% Number Al Ti Mn Co Zr of spectrum 3.91 94.89 1.20 1 \_ 2 0.90 36.49 10.45 52.15 3 41.20 \_ 3.36 1 54.444 1.10 44.04 1.06 53.80 \_ 5 0.34 39.16 \_ 4.28 56.22 0.99 6 41.60 \_ 1.98 55.427 0.31 20 9.57 70.13 \_ \_ \_ 3.98 94.94 8 1.08 9 4.10 94.19 0.71 \_ 1 4.24 0.36 1.68 10 90.35 3.37 1.78 65.86 11 0.51 6.43 25.420.77 \_ 12 45.65 4.10 49.49 13 0.77 42.01 -1.70 55.53

Table 2. Chemical composition of brazed joint (OT4 base metal,

even at short indicated holding. At that weight fraction of titanium and iron does not change in the seam cross section, whereas zirconium weight fraction is somewhat increased in the seam center. This can be explained by formation of zirconium solid solution in titanium (Figure 3).

4.10

3.87

3.19

14

15

16

92.77

88.57

92.94

0.32

0.27

2.18

\_

1.70

2.82

7.29

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Results of X-ray microspectrum analysis of metal of seam and fillet area are represented in Table 2 and Figure 4 in more details.

First of all, it should be noted that composition of the base metal, determined for selected area, completely corresponds to the requirements of standard for OT4 alloy (Table 2, spectra 1 and 8).

Measurement results obtained in cross section of the seam in some distance from the fillet (Table 2, spectra 14 and 15) are close to these values, i.e. chemical composition of the seam is close to composition of metal being brazed even with that holding at brazing temperature. Concentration of titanium and aluminum are virtually corresponds with the same for brazed metal (see Figure 4, Table 2).

Data of chemical composition of the fillet metal (Table 2, spectra 2–7), regardless some differences, show general tendencies, namely significant reduction of titanium and aluminum content, high content of zirconium and alternating content of cobalt (1.60–9.57 %) in comparison with metal of the seam. At that spectrum 7 differs from the rest by particularly low content of ti-



Figure 4. Microstructure and areas of X-ray microspectrum analysis of brazed joint (OT4 base metal, Ti-Zr-Co BFM)

tanium and high content of zirconium. This can be explained by the fact that the fillet has double-phase structure, and result depends on that what phases were in probe zone. It can be added that the base metal adjacent to the seam has typical plate structure consisting of two phases.

Mentioned above is confirmed by investigations of distribution of elements in base metal to fillet interface (see Table 2, spectra 9-12). Chemical composition in specified area (spectrum 11) is close to composition of the base metal, further clear appearance of tendency mentioned above is observed, i.e. reduction of titanium and aluminum content and increase of that for zirconium. Spectrum 13 corresponds to the fillet composition in full.

Distribution of elements in the seam metal has no principal differences from considered above, except for iron, concentration of which smoothly increases in the central seam area when using BFM of Ti-Zr-Fe system.



Figure 5. Microstructure and near-fillet areas of X-ray microspectrum analysis of brazed joint (OT4 base metal, Ti-Zr-Fe BFM)





**Figure 6.** Tensile strength of brazed butt  $\sigma_t$  (1) and lap  $\tau_t$  (2) specimens (base metal OT4) obtained using BFMs on the basis of systems: 1 - Ti-Zr-Co; 2 - Ti-Zr-Cu-Ni; 3 - Ti-Zr-Cu-Ni; 4 - Ti-Zr-Fe; 5 - Ti-Zr-Fe

Base metal adjacent to the fillet has doublephase structure (Figure 5), its chemical composition is close to initial composition of OT4 alloy and includes insignificant quantity of chemical elements of the BFM (Table 3, spectrum 1).

Significant reduction of content of titanium, aluminum, manganese and substantial increase of iron and zirconium are observed in the interface (see Table 3, spectrum 2). This tendency is less apparent in the fillet area (Table 3, spectrum 3).

It can be concluded considering mentioned above that 15 min holding is enough for formation in the seam of alloy close to metal being brazed during brazing using both BFMs. At that eutectic structure is preserved in the fillet.

Series of butt and lap specimens from titanium alloys OT4 and VT6 was manufactured using standard and investigated BFMs (Table 4) in order to get an idea about strength characteristics of brazed titanium alloys.



Figure 7. Tensile strength of brazed butt  $\sigma_t$  (1) and lap  $\tau_t$  (2) specimens (base metal VT6) obtained using BFMs on the basis of systems: 1 - Ti-Zr-Co; 2 - Ti-Zr-Cu-Ni; 3 - Ti-Zr-Cu-Ni; 4 - Ti-Zr-Fe; 5 - Ti-Zr-Fe

Table 3. Chemical composition of brazed joint (OT4 base metal, Ti–Zr–Fe BFM), wt.%

Number of spectrum	Al	Ti	Mn	Fe	Zr
1	2.93	89.95	0.89	3.57	2.67
2	1.70	76.57	0.41	10.73	10.59
3	0.58	48	0.41	21.67	29.34

Table 4. BFMs and brazing modes of wrought titanium alloys OT4 and VT6  $\,$ 

BFM system	Initial state of BFM	Brazing temperature, °C
Ti–Zr–Co	Cast	920
Ti–Zr–Cu–Ni	Amorphous strip	1000
Ti–Zr–Cu–Ni	Cast	1000
Ti–Zr–Fe	Cast	990
Ti–Zr–Fe	Amorphocrystalline strip	990

Analysis of results of mechanical test showed that proposed alloy systems provide for mechanical properties of the brazed joints at the level of that obtained during brazing using known BFM of Ti–Zr–Cu–Ni system. This is achieved at significantly lower brazing temperature applying BFM of Ti–Zr–Co system. Figures 6 and 7 show the results of tests (average of three measurements).

Application of BFM of Ti–Zr–Fe system in brazing of intermetallic alloy Ti–45Al–2Nb–2Mn + 0.8 vol.% TiB<sub>2</sub> provides formation of the seams of alternating width with two-phase structure ( $\gamma$ -TiAl and Ti<sub>3</sub>Al) containing no eutectic constituent (Figure 8).

Width of the seams and their chemical composition are determined by capillary peculiarities of the BFM and diffusion processes taking place at liquid BFM-solid substrate interface in brazing. Formation of the seam with plate (lamellar)



**Figure 8.** Microstructure of brazed joint of titanium aluminide obtained using BFM of Ti–Zr–Fe system with cast structure



structure close to the base metal structure [12] is observed. Chemical composition in these areas is virtually similar to the base metal. The latter preserves lamellar structure after brazing.

Results of the investigations obtained with the help of electron scanning microscopy and Xray spectrum microanalysis show that chemical composition and structure of the seam metal significantly differ from such for initial BFM. This is caused by gradient of concentration of constituent elements of BFM and base material at the phase interface, capillary (0.05 mm) gaps and non-equilibrium conditions of solidification. Diffusion processes taking place at the phase interface of solid material with liquid BFM, in particular, result in leveling of aluminum concentration in the base material and seam metal and formation of phases with aluminum concentration corresponding to such for the base material.

Similar formation of the seams takes place in brazing using BFM of Ti-Zr-Mn system. There are areas where seam metal has plate (lamellar) structure close to structure of the base metal. Intergrown grains of the base material are observed in some areas, and chemical composition of the metal at the joint interface is virtually identical to the base metal.

Results of strength tests carried out at room temperature using butt specimens showed that alloys based on Ti-Zr-Fe and Ti-Zr-Mn systems provide for brazed joints of 650–700 MPa tensile strength, and this strength is at the level of shortterm strength of material being brazed. Strength of the brazed joints makes around 300 MPa at 700 °C testing temperature.

Results of creep-rupture tests conforming working capacity of the joints under conditions of maximum approximation to service ones [13] are an important factor of high-temperature strength of the brazed joints. Brazed specimens did not failure in a course of 500 h during creeprupture tests at temperature 700 °C and 140 MPa stress. Increase of stress up to 200 MPa did not cause specimen failure.

It should be noted based on test results that strength of the brazed joints obtained using BFM of Ti–Zr–Cu–Ni system is 12–18 % lower than at application of BFMs of Ti-Zr-Fe and Ti-Zr-Mn system.

Thus, BFMs, developed on the basis of performed investigations, allowed obtaining brazed joints of intemetallic alloy  $\gamma$ -TiAl close on structure and properties to the base metal. Received results can be a foundation for development of new critical structures of different designation from new perspective titanium materials based on intermetallics using BFMs considered above. Developed BFMs contain no copper or nickel and can be used for the parts of technical as well as medical designation.

#### **Conclusions**

1. Brazing filler metals covering temperature range of brazing of modern structural titanium materials based on solid solutions, as well as intermetallics, were proposed as a result of complex investigations of alloys of Ti-Zr-Fe, Ti-Zr-Mn and Ti-Zr-Co systems.

2. Brazing of alloys based on solid solutions using indicated BFMs showed the strength characteristics at the level of that obtained using known BFMs, even if they are received at lower brazing temperature.

3. Results of mechanical tests of the brazed joints from  $\gamma$ -TiAl intermetallic based alloy showed that proposed BFMs provide fullstrength of the base material at room and elevated temperature, as well as at creep-rupture tests.

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### INFLUENCE OF WELDING HEATING ON FATIGUE STRENGTH OF HOLLOW STRUCTURES FROM HIGH-STRENGTH FINE-GRAINED STEELS

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Welding trials and technological assessment of square hollow sections of  $80 \times 80 \times 10$  mm size from steel of SG69Q grade were performed. Welds were assessed on macrosections, by hardness measurement, their static bending and impact testing. In order to determine fatigue strength at bend testing with cycle asymmetry factor R = 0.5 under the conditions of rail transport works, these sections were joined by mechanized gas-shielded solid and flux-cored wire welding. Comparison of section testing results and their compliance to normative requirements was performed. Recommendations on application of various sections in structures with cyclic loading are given. 18 Ref., 5 Tables, 12 Figures.

**Keywords:** arc welding, metal electrode, shielding gas, high-strength steels, hollow sections, fatigue strength, light-weight structures, recommendations for application

Such structural elements as square and rectangular hollow sections have been applied in mechanical engineering and instrument making, in vehicle manufacture and steel structure fabrication for dozens of years. Seamless hot-rolled hollow sections without a longitudinal weld are exactly the elements with uniform material properties over the entire cross-section. As the profiles are not subjected to cold deformation in manufacturing, they feature excellent weldability, particularly in the vicinity of the edges. Moreover, seamless hot-rolled hollow sections have a wide range of wall thicknesses, even at a comparatively small edge length. Owing to their large flat contact surfaces as a result of small angular rounding-off, the sections are highly suitable for compact structures exposed to high loads [1, 2].

Steels of 355 MPa strength class are predominantly used for steel structures. Higher strength steels are needed for agricultural machinery, where square and rectangular sections with up to 500 MPa yield limit are used (Figure 1). In addition, high-strength steels with up to 960 MPa yield limit, proposed by Vallourec & Mannesmann, already begin to be applied in transportation engineering that became an essential contribution to solving the task of reducing structure weight.

However, application of high-strength steels in transportation engineering involves the need to allow for fatigue properties of the joints. In



**Figure 1.** Production of square MSH-sections in the last rolling stand (*a*), and their application in agricultural machinery engineering (*b*)

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terms of admissible number of load cycles at a specified loading pattern, evaluation of metal properties, primarily in the weld area, is highly important.

In this connection investigations were conducted to assess the loading impact on properties of welded parts of a certain configuration. Selection of dimensions, steels, spectrum and scope of investigations is oriented towards the requirements of transport engineering.

The subject of discussion is rectangular or square hollow steel sections of MSH grade, manufactured from unalloyed or fine-grained structural steels by Vallourec & Mannesmann in compliance with EN 10 210–1 [3]. Supply spectrum includes hot-rolled sections of  $40 \times 40$  to  $300 \times 300$  mm size (square) and those of  $50 \times 30$ to  $300 \times 200$  mm size (rectangular). In addition, MSH-sections can be manufactured with longitudinal welds made by high-frequency induction resistance welding, and MSH-sections of  $400 \times$  $\times 400$  (square) or  $500 \times 300$  mm size (rectangular) from hot-rolled metal can be produced. Wall thickness is standard — up to 20 mm.

Improved MSH-sections developed on the basis of the concept of alloying fine-grained steels of the strength from 690 to 890 MPa are put on the market under the name of FineXcell<sup>®</sup>. They have minimum yield limits of 690 MPa in the lower range of wall thickness ( $\geq 16$  mm) and good toughness.

Welding of fine-grained structural steels. In welding it is necessary to make sure that material properties will not deteriorate more from thermal impact than) from the load of the structure proper. As welding conditions essentially influence the joint properties, the task of specialists is to ensure the required metallurgical properties of the deposited and HAZ metal. Mechanical properties of welded joints are determined by temperature-time welding cycle. Current, voltage, welding speed, as well as item thickness and weld geometry, have a great influence. These parameters proper determine the temperaturetime cycle of welding, often characterized by metal cooling rate in the temperature range 800-500 °C. High rate of metal cooling from the austenitic region can lower the toughness of welded joint HAZ metal. More over, the risk of cold cracking in the deposited and HAZ metal can become higher. As a result of slower cooling of the weld, its strength can decrease and it will no longer correspond to strength properties of base metal.

An effective means of cold crack prevention is preheating. Preheating temperature is understood to be item temperature in the vicinity of the weld directly before welding. Preheating temperature is assigned, depending on the material, its wall thickness, weld geometry and heat input value. Temperature is increased with increase of metal thickness that slows down cooling of weld metal zone and promotes hydrogen evolution. In addition, preheating has a positive influence on inner stressed state of the joints. Steel proneness to cold cracking has an essential influence on welding operations cost. For this reason it is important to classify steels, depending on their crack resistance. Useful information on this problem is published in [4–7], where carbon equivalent CET (%) is proposed as crack resistance criterion:

CET = C + (Mn + Mo) / 10 + (Cr + Cu) / 20 + Ni / 40.

At appearance of cold cracks the situation can be such that preheating temperature was selected correctly, but the actual heat removal in the item was incorrectly assessed. First, according to [8], preheating temperature should be measured at a sufficient distance from the weld. Secondly, it is necessary to thoroughly heat the locations where several welds meet, and where, alongside a more intensive heat removal, a 3D stressed state can be in place, further promoting cold cracking. In addition to chemical composition of base and deposited metal, appearance of cold cracks largely depends also on wall thickness, hydrogen content in the deposited metal, as well as inner stressed state of the joint. Method of calculation of minimum preheating temperature, known as CET concept, is included into «STAHL-EISEN-Werkstoffblatt SEW 088» journal [9], as well as into EN 1011 standard [10].

Heat input Q is calculated by determination of energy input E and depending on thermal efficiency of the process  $\eta$  by a known equation described in [10]:

$$E = \frac{U_{a}I_{w}\cdot 60}{v_{w}\cdot 10000} \text{ (kJ/mm) or } Q = \frac{\eta U_{a}I_{w}\cdot 60}{v_{w}\cdot 10000} \text{ (kJ/mm),}$$

where  $U_{\rm a}$  is the arc voltage, V;  $I_{\rm w}$  is the welding current, A;  $v_{\rm w}$  is the welding speed, cm/min;  $\eta = 0.85$  is the MAG welding thermal efficiency.

**Technological investigations.** Selection of welding consumables. For welding high-strength fine-grained steels the filler materials of various manufacturers and suppliers are selected, depending on their yield limit. Both solid and flux-cored wire is used, in keeping with EN DIN 18276, in shielding gas atmosphere M21 ( $CO_2 + 82 \%$  Ar), according to EN ISO 14175 [11]. When selecting the wire, it is necessary to take into account not only its cost, but also fabrication costs for the produced structure.



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In this respect, seamless rutile, basic or fluxcored wires with metal core have advantages, compared to solid wire, as they provide specific properties. The following filler consumables were used in welding experiments described below:

• to EN ISO 16834-A solid wire GMn4Ni1.5CrMo [12] (ED-FK 800);

• according to EN ISO 17632-A T69 6 fluxcored wire with metal powder for root welding MnN2NiCrMo MM1H5 (STEIN-MEGAFIL 742 M) [13], as well as to EN ISO 17632-A T69 6 ZPM1H5 rutile flux-cored wire for filling and facing beads (STEIN-MEGAFIL 690 R).

The above materials are suitable for welding fine-grained structural steels, in particular the studied MSN-profiles. In this work the influence of these consumables on weld geometry and longterm strength of welded joints was assessed.

Experimental procedure. Requirements to welded joint quality are made, first of all, at application of these hollow sections in structures with a dynamic load [14, 15]. Hollow sections were welded both for determination of mechanical and technological characteristics, and for fatigue strength assessment. In the latter case, dimensions and geometry of welds in welded assemblies were determined. 36 assemblies were made, where a hollow section of 150 mm length was inserted between the joined sections (Figure 2). Solid and flux-cored wire were used to weld 18 samples each. Then welding parameters were determined, depending on the groove. In parallel, fixtures for tack welding and welding were developed, and welding sequence was determined. During technology development preheating temperature and thermal mode was calculated. Preheating and interpass temperature of 80-100 °C was assigned, and M21 gas was used as shielding atmosphere.

The process consisted of mounting the parts in the fixture for alignment, positioning and tack welding of sections, grinding the tack welds be-



Figure 2. Schematic and dimensions of samples

fore root welding, welding the root bead in the downhand position alternatively from two sides, mounting the parts in the fixture for welding of intermediate beads (HV-welds when making a fillet weld (Figure 3, a)), upper bead welding (HV-weld when making the butt weld). Finished welded joints, made for fatigue strength studies, are shown in Figure 3, b.

*Edge preparation and welding sequence*. In order to prepare samples for welding for future evaluation of joint fatigue strength, the sections were machined so that in the vicinity of the weld the angle of opening was 45° at root face height of 1 mm. A gap of 3 mm was made for reliable welding of weld root; joint geometry and welding sequence are given in Figure 4.

**Testing methods.** *Macrosections*. Non-destructive testing (NDT) was performed to assess welded joint quality. Then macrosections were cut out of butt and fillet welded joints (Figure 5).

*Tensile testing*. This kind of testing should show the correspondence of yield limit and relative elongation of the joints to requirements made



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Figure 3. Welded assemblies with fillet (a) and butt (b) welds

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**Figure 4.** Schematic of edge preparation and sequence of welding for fillet (*a*) and butt (*b*) weld: W - weld root; ZWL - intermediate layer; DL - upper bead

of hollow sections from FineXcell<sup>®</sup> 690ImpactFit50 with a certain wall thickness.

Test results presented in Figure 6 showed that both the butt and fillet joints meet the requirements made of them. Tensile strength of the deposited and base metal was higher than the normative requirements in all the joints.



Figure 6. Experimental data on static strength testing of SG69Q steel welded joints

*Toughness*. Table 1 gives the requirements to FineXcell<sup>®</sup> 690ImpactFit50 material. As is seen from Figure 7, both in the HAZ and deposited metal the required impact energy (23 J) in the longitudinal direction of the butt joint was achieved at all the testing temperatures, in keeping with EN 10025–1 and 10025–6.

Hardness characteristics. In samples, shown in Figure 8, hardness was measured (HV10) from the side of weld surface and from its lower side. Comparison of measured hardness values of metal deposited by flux-cored and solid wires showed lower hardness for flux-cored wire. Maximum values were observed in HAZ metal. Series of hardness measurements from the weld upper side demonstrated higher values — this comparison



Figure 5. Macrosections of butt and fillet welds made by solid M1S, M1K (a, b) and flux-cored F1S, F1K (c, d) wires

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Joint	Minimum impact energy, J, on three Charpy samples at control temperature, °C				
direction	-50	-40	-20	0	
Across	27	30	40	65	
Along	16	27	30	40	

Table 1. Impact toughness requirements made of hollow sections from FineXcell\* 690ImpactFit50 material with wall thickness  $\geq\!20~mm$ 



Figure 7. Impact energy of  $80 \times 80 \times 10$  mm samples from SG69Q steel

is given once more in the diagram (Figure 9). It is seen that in samples welded with flux-cored wire hardness values are lower both in the deposited and in the HAZ metal.

*Fatigue strength.* Let us consider the approach and results of fatigue testing of welded joints of  $80 \times 80 \times 10$  mm sections from SG69Q steel, made taking IIW recommendations [16] into account. It was necessary to study the types of loads given in Table 2, allowing for the applied filler materials.

To determine the fatigue strength, Woehler testing was performed. About 7-8 full-scale samples were made for each Woehler line, which were tested for four-point bending (Figure 10).

In order to compare the steady-state fatigue strength with IIW recommendations [16], testing with a constant coefficient of cycle asymmetry R = 0.5 was required. Then comparison was performed, allowing for reliability probability  $P_u =$  97.5 % with randomly taken scatter band  $T_s$  ( $P_u = 90 \%$ : $P_u = 10 \% = 1:1.5$  [17]), first in the form of tolerable stress. After that recalculation to the range of alternating stress  $\Delta S$ , allowing for the coefficient of cycle asymmetry, was performed for comparison with IIW recommendations. Samples without tearing, tested up to stress of  $N = 5 \cdot 10^6$  cycles, were called random and were not taken into account at evaluation.

In order to determine the normal nominal stress, moment of resistance of rolled section  $W_{\rm el} = 53.5 \text{ cm}^3$  [18] was assumed as the charac-



**Figure 8.** Hardness measurement in fillet (*a*) and butt (*b*) welds made with flux-cored F1K, F1S and solid M1K, M1S wires



Base metal HAZ Weld metal HAZ Base metal Figure 9. Hardness measurement on weld upper side: 1 -flux-cored wire F1K - fillet weld; 2 -flux-cored wire F1S - butt weld; 3 -solid wire M1K - fillet weld; 4 -



Figure 10. Real testing process for 4-point bending

Table 2. Sample schematic and kinds of load



Table 3. FAT 50 IIW classification [16] for HV-weld when making fillet weld on sample No.424

Structural detail	Description	I	FAT	Remark
	Description		Aluminium	Tenlark
	Splice of rectangular hollow section, single-sided butt weld, potential failure from toe Wall thickness >8 mm Wall thickness <8 mm	50 45	20 18	NDT of welds in order to ensure full root penetration

teristic of welded joint cross-section with bending load after control testing in a real section. Bending moment M required furtheron was calculated by geometrical conditions of four-point bending, in keeping with the theory of elasticity.

As is seen from Figure 11, the point of crack initiation in all the samples is located predominantly in the corners at transition to the weld, in connection with a higher stress concentration, associated with higher rigidity of the part in this area.

Figure 12 gives a comparison of experimental results with the data of IIW recommendations. As was noted above, IIW recommendations express fatigue strength by Woehler lines. FAT strength class implies the admissible range of nominal stress  $\Delta S$  [MPa] in the case of an undercut at a stationary stress cycle  $N = 2 \cdot 10^6$ . Here the slope of steep-falling portion of Woehler curve is assigned as m = 3 for welded joints.

Some results are more obvious at further comparison of samples with undercuts in keeping with IIW recommendations. In both the joints the slope of Woehler curve is m = 3. HV-weld at performance of fillet weld (HV-K-M, HV-K-F) can be tentatively included into fatigue strength class FAT 50 (Table 3). The formed circumstances are even more unfavourable, as here, in view of the lower rigidity of the hollow section placed in-between, additional stresses are applied to the corner regions (compare tears in Figure 10).



Figure 11. Cracks at weld transitions in section corners



Figure 12. Fatigue curves for joints made with solid (a) and flux-cored wire (b)

During performance of butt weld (HV-R-M, HV-R-F) the HV-weld can be included into FAT 56 class, if we assume that NDT confirmed the appropriate quality of weld root (Table 4). Table 5 gives comparison of NDT results with IIW data.

Although the slope of all the derived Woehler lines is flatter than in IIW recommendations,

Table 4. FAT 56 IIW classification [16] for HV-weld when making butt weld on sample No.234

Structural detail	Description	Fatigue	class (FAT)	Domork
Structural detail	Description	Steel	Aluminium	i cinar k
	Transverse butt weld splice in rec- tangular hollow section, welded from one side, full penetration, root crack Root NDT	56	25	Welded in flat position



56

Table 5. Comparison of experimental results with IIW data

IIW requirements	$\Delta S$ , MPa	m	Sample marking	$\Delta S$ , MPa	m
FAT 50	50	3	HV-K-M	64	3.4
			HV-K-F	66	3.5
FAT 56	56	3	HV-R-M	78	3.3
			HV-R-F	91	4.0

there is no point of crossing comparable Woehler FAT-lines, because of higher values of vibration strength at  $N = 2 \cdot 10^6$  cycles.

Fatigue strength should not be overestimated either in the low-cycle region (cranes, gantry rails, pressure apparatuses), or in regions with high-cycle loads. It should, however, be borne in mind that in this case joints of hollow section assemblies (SG69Q) were evaluated, and calculation was performed according to FAT classes by IIW recommendations.

Therefore, application of high-strength improved fine-grained structural steels in many cases allows reducing wall thickness and, thus, lowering material and processing costs. A wide strength range enables limitation of the dimensions and weight of the part, allowing for the conditions of production and loads. Only application of these steels allows overcoming the established limits in some cases. Performed welding experiments are indicative of usability of the applied sections. With the respective edge preparation and observation of technological rules, it is possible to perform sound joints with good mechanical properties. Fatigue testing confirmed the applicability of the described sections also for structures with cyclic loads. Following the respective norms, instructions and recommendations is a necessary condition.

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### PECULIARITIES OF EXPLOSION WELDING OF STEEL WITH CAST IRON

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Products of cast iron find a wide spreading in modern machine building, in particular for manufacture of friction discs, in which the plates of cast iron are fastened by bolts to steel or copper plates. According to hypothesis about formation of impact-compressed gas (ICG) and its thermal ionization at the interface with the formation of thin layers of low-temperature («cold») plasma in a welding gap during explosion welding, the conditions of formation of carbon steel–low-alloy cast iron joint were determined. Conditions of explosion welding and heat treatment were set for producing full-strength joint and prevention of defects in the form of cracks and spallings along the entire line of contact between the steel and cast iron. The carried out evaluation calculations of parameters of ICG with account for super-speed flowing of surface being welded by it showed that change in rate of contact spot from 2400 up to 4000 m/s leads to increase in temperature and decrease of length of ICG region, as well as time of plasma effect. 6 Ref., 1 Table, 1 Figure.

**Keywords:** explosion welding, welding gap, impactcompressed gas, gas parameters, cleaning, contact spot

In manufacture of friction discs the cast iron plates are fastened to steel or copper ones using bolts. Replacement of the bolted joint by strong welded joint along the entire surface of the cast iron disc will allow increasing the strength of the product, its manufacturability, making possible to apply welding in manufacture of cast iron products. However, the bolted joint does not guarantee the dense contact along the entire surface, thus deteriorating the heat removal from the cast iron disc and contributing to its non-uniform heating. The presence of local zones of preheating leads to cracking and pouring out of cast iron.

The main drawbacks of cast iron as a structural material are its low ductility and poor weldability by all the methods of welding, including explosion welding. The experience of previous works on producing of steel + cast iron bimetal by explosion welding showed that due to a low ductility the defects in the form of cracks, spallings, delaminations are formed in the process of welding [1].

During development of the technology of explosion welding of cast iron with steel it is necessary to solve the following main tasks:

• producing of strong joint over the entire surface;

• prevention of formation of cracks and fractures of cast iron during welding; • investigation of mechanical properties and structure of produced joints, as well as effect of postweld heat treatment on them.

To solve the put tasks, the following methodology was developed, which provided on the basis of published data and experience of production of bimetals by explosion welding:

• evaluation of feasibility of explosion welding of steel with cast iron;

• working out of technological solutions, reducing the probability of formation of cracks and fractures of cast iron during welding;

• development of pilot technology of producing steel + cast iron bimetal and investigation of produced joints.

At present, a large experimental and theoretical information was accumulated on the problem of formation of joints in explosion welding, which was generalized in works [2, 3]. Some hypotheses were made, explaining the formation of joints from different points of view. In the zone of collision in explosion welding the high pressures are developed, an intensive plastic deformation is going on, accompanied by significant increase in temperature of metals in the collision zone.

The explosion welding is characteristic of three stages of process for formation of strong bonds between the atoms of metals being joined, realizing in the following sequence: cleaning and activation of contact surfaces, formation of physical contact, and volume interaction. The quality of explosion welding is defined, first of all, by the processes, proceeding in a welding gap ahead of the contact spot [4], i.e. by cleaning and activation of surfaces being joined.

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Evaluation of parameters of impact-compressed gas (ICG) in the welding gap was made by procedure [4] for the following explosion welding mode parameters: rate of contact spot  $v_c$  is varied from 2400 up to 4000 m/s, ratio of mass of explosive to mass of flyer plate — from 1.2 to 0.7, welding gap — from 8 to 1 mm, Mach number is varied from 7 up to 11.6. The calculations showed that the ICG parameters in welding gap in this case are varied in the following limits: pressure — from 7 up to 13.7 MPa, temperature — from 2500 up to 5000 K.

The carried out evaluation calculation by the procedure [4] showed that increase in rate of contact spot improves the conditions of cleaning and activation of surfaces being welded due to increase in rate and temperature of ICG in welding gap and parameters of plasma at the interface between the ICG and surfaces being welded. Time of plasma effect t can be determined by formula  $t = 1/v_c$ , where l is the length of ICG region, determined by procedure [4]. With increase in rate of contact spot from 2400 up to 4000 m/s, the time of plasma effect is decreased from  $2.4 \cdot 10^{-5}$  to  $1.12 \cdot 10^{-5}$  s.

Increase in rate of explosive detonation will lead to the increase in speed and energy of collision of flyer plate with the base. This, in turn, will increase the plastic deformation of the base sheet and probability of appearance of cracks and fractures in it. It is possible to control the speed of plate flying by changing the welding gap. The procedures of calculation of collision speed, described in works [2, 3, 5], does not take into account the value of welding gap and it was only suggested in work [6] to determine the angle of collision and speed of flying  $v_0$  with account for value of the welding gap:

for mixed explosives

$$v_0 = 2D \sin \frac{0.49r}{r + 2.71 + 0.184/h};$$
 (4)

for ammonite

$$v_0 = 2D \sin \frac{0.416r}{r + 2.71 + 0.184/h},$$
 (5)

where D is the detonation rate of explosive; r is the dimensionless parameter equal to ratio of explosive mass to mass of the flyer plate; h is the gap height.

Calculation of flying speed and angle of collision by these formulas for the above-given explosion welding conditions and at varying the welding gap from 1 up to 8 mm showed that flying speed at decrease of welding gap from 8 to 1 mm is 2 times reduced. This will allow 4 times decreasing the collision energy and, as a consequence, decreasing the probability of appearance of cracks and fractures in cast iron.

Thus, to guarantee the strong joints in explosion welding of cast iron with steel, the process should be performed under conditions with increased rate of detonation of 3500-4000 m/s and 2-1 mm welding gap to prevent the formation of cracks and fractures of cast iron.

As initial material in conductance of experimental investigations, the sheet of steel of 08kp grade (killed) of  $4 \times 300 \times 500$  mm sizes and cast plates of ferrite-pearlite low-alloy cast iron with a lath graphite (further — cast iron) of  $8 \times 180 \times 350$  mm sizes were used. A parallel scheme of explosion welding was applied. Two explosion welding conditions were used:

• rate of detonation of explosive D = 2400-2500 m/s; dimensionless parameter r = 1.2; welding gap is 8 mm that provides the flying speed of 500-550 m/s; calculation angle of collision is 8-9°;

• D = 3500-3700 m/s; r = 0.7; welding gap is 1.8 mm that provides the flying speed of 340–400 m/s; calculation angle of collision is 2–3°.

It was found as a result of experiments that in welding using the first variant the cutting of overhanging of steel plate, significant deformation of plate of cast iron with its partial fracture and full absence of the joint over the entire surface of cladding with prints of relief of mechanical treatment of cast iron plate are observed. The absence of adhesion spots on the surfaces being welded indicates that the preset condition of explosion welding did not provide cleaning and activation of cast iron surface ahead of contact spot.

Strength of steel + cast iron joint and hardness of cast iron\* depending on heat treatment

Heat treatment	Tensile strength of joint σ <sub>t</sub> , MPa	Hardness of cast iron <i>HB</i>	Hardness of white phase HB
Without heat treatment	130-135	220	514
Heating up to 550 °C, holding from 4 up to 6 h, furnace cooling	145-150	175	322
Heating up to 700 °C for 1 h, furnace cooling	155-160	160	No white phase
*			

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 $\sigma_t$  of initial cast iron is 200 MPa, hardness HB is 163–219, no white phase.

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Microstructure of steel 08kp + alloyed cast iron joint: a – without heat treatment; b – heat treatment at 550 °C for 4–6 h, furnace cooling; c – same at 700 °C for 1 h, furnace cooling

Experiments using the second variant showed that the joining of steel with cast iron occurs over the entire surface with a negligible deformation of welded plates. The ultrasonic testing of some two-layer plates revealed small  $(2-3 \text{ cm}^2)$  angular lacks of penetration.

Tensile tests of the joint for separation of a cladding layer were carried out on specimens without heat treatment and specimens with two different heat treatments. The first condition was a heating up to 550 °C, holding for 4–5 h and furnace cooling; the second — heating up to 700 °C for 1 h and furnace cooling (Table). The tests showed the joint strength after explosion welding was 130–135 MPa, heat treatment by the first condition increased the joint strength up to 145–150 MPa and decreased its hardness. The strength of joint of specimens, heat treated by the second condition, was 155–160 MPa.

The microstructure of joint of specimens without heat treatment showed the presence of a white phase (Figure, a), hardness of which was HV 514 (see the Table). Chemical composition of the white phase corresponds to that of cast iron. Examination of microstructure of joint of specimens after heat treatment (550 °C, 4–6 h, furnace cooling) showed the decrease in sizes and amount of the white phase (Figure, b), which hardness was HV 322. Microstructure of joint of specimens after heating to 700 °C for 1 h and furnace cooling showed the absence of the white phase (Figure, c), consequently, this heat treatment condition allows annealing the inclusions of the chilled cast iron (white phase) at the joint interface, thus increasing the mechanical properties of the joint.

Metallographic examination of steel 08kp + cast iron joint showed that the joint has no large regions of cast inclusions. The cast iron cracking was not revealed, which could be formed in producing bimetal, thus proving the favourable effect of small welding gaps and applying the explosive with increased rate of detonation.

The analysis of investigation of microhardness of specimens showed that hardness is levelled

and strength of the joint is increased in the zone of joint at heat treatment by the condition of cast iron annealing (700  $^{\circ}$ C for 1 h and furnace cooling).

#### Conclusions

1. Decrease in welding gap from 8 to 1.8 mm leads to reduction in rate of collision of the flyer plate with base almost 2 times and, consequently, the collision energy is 4 times decreased.

2. Experiments on explosion welding of steel with cast iron confirmed the evaluation calculations and showed that increase in rate of contact spot with simultaneous decrease of parameter r and welding gap to 1.8 mm creates conditions of producing full-strength joint between the carbon steel and low-alloy cast iron having no deformations and defects (cracks) on the brittle cast iron surface.

3. Investigation of microstructure of joint of steel with cast iron before and after heat treatment showed that chemical composition of inclusions of a white phase corresponds to that of low-alloy cast iron, i.e. it is a chilled cast iron. Heat treatment in the condition of annealing provides the transformation of structure of the chilled cast iron into the initial one.

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## CALCULATION OF MODE PARAMETERS OF WALL BEAD DEPOSITION IN DOWNHAND MULTI-PASS GAS-SHIELDED WELDING

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Application of narrowed grooves is an effective method to increase welding process efficiency at simultaneous reduction of financial and energy expenses. One of the main difficulties in implementation of multi-pass welding technology 10 mm thick rolled metal structures using automated and robotized units is a high possibility of defect formation such as lack of penetration, in particular, in layer of the first pass (wall bead). Control of welding parameters determining heat input is one of the methods of penetration regulation. Dependencies between welding parameters in problem areas of the grooves and penetration depth were determined experimentally. On their basis the equations for determination of parameters of bead deposition at given coefficient of areas were obtained. These results became a part of algorithm development to control welding mode parameters in automated and robotized units. 12 Ref., 1 Table, 6 Figures.

**Keywords:** mechanized consumable electrode welding, total heat efficiency of welding process, full-factor experiment, coefficient of areas, mathematic modelling, determination of parameters of welding mode

Increase of welding process efficiency is still a relevant problem under conditions of current welding production. Application of narrowed grooves [1] is one of the proficient methods allowing significantly increasing welding process efficiency as well as reducing financial and energy expenses. Along with it the efficiency increase requires implementation of automatic and robotized systems. However, in realization of multi-pass consumable electrode welding technology on automatic and robotized units some difficulties appear, related to defect formation such as lack of penetration. In particular in wall bead deposition [2], in order to obtain the weld metal with needed mechanical properties, favorable structure, minimal deformation and required weld shape, as well as lower possibility of hot and cold cracks formation [3] it is necessary decreasing heat input. Under conditions of decreased heat input in multi-pass welding due to increased heat abstraction to base metal the possibility of lack of fusion between bead and groove edges, as well as the nearby beads, considerably grows.

One of the regulation methods of the base metal penetration in groove problem areas is changing of process parameters (current, voltage, speed of welding), that determine heat input and, therefore, weld formation conditions, in particular, depth of edge penetration [4, 5]. Work [6] covers the study results of welding mode parameters influence (reversed polarity welding current, arc voltage, distance between electrode tip and edge) on wall bead shape in submerged arc welding. It was determined that the parameter, which defines mechanical jamming of slag crust, is an angle of transfer of wall bead surface to the edge. The main parameters of mode determining its value are the arc voltage, distance between electrode tip and edge as well as welding speed. However, this study does not cover the questions of preparation angle influence on lack of penetration near bevel edge.

There is also a well-known method, when to prevent the lack of penetration between bead and bevel edge, the automatic welding with weaving and increased pulsed current towards edge being welded is used. However, determination of accurate value of pulse power is necessary for that (due to evaluation of efficiency use of arc heat energy).

Thus, issues regarding determination of bevel angle influence on bead formation, optimization of groove parameters, as well as modes of setting of wall bead deposition with given area coefficient *k* equal ratio of penetration area to deposition area  $F_{\text{pen}}/F_{\text{d}}$ , which guarantees absence of lack of penetration, are still relevant.

The aim of the present work is the determination of dependencies between parameters of wall bead deposition in mechanized multi-pass gas-shielded welding and penetration depth, as well as definition of modes of wall bead deposition with set area coefficient based on obtained dependencies.

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Figure 1. Scheme of bead passing

Efficient control of penetration of metal being welded under conditions of arc welding is possible only having information about main dependencies of this process, as well as qualitative and quantitative effect of welding mode parameters on size and shape of penetration zone. Statistically recognized dependencies between energy parameters of welding mode (welding current, arc voltage, welding speed etc.), on the one hand, and parameters, characterizing quality of welded joint, on the other hand, make a basis of the most mathematical models developed for welding process control. Depth of penetration of welded metal and area of weld section [7, 8] refer to the parameters determining weld geometry and quality of welded joint. Efficiency of application of arc heat energy for formation of welded joint is evaluated using total heat efficiency of welding process  $\eta_w$ , which determines the ratio of conditional heat input of weld metal being fused per time unit to heat power of welding heat source [5]:

$$\eta_{\rm w} = \frac{v_{\rm w} F_{\rm weld} \gamma_{\rm m} H_{\rm melt}}{Q} =$$

$$= \frac{v_{\rm w} (F_{\rm weld} + F_{\rm pen}) \gamma_{\rm m} H_{\rm melt}}{Q} = \eta_{\rm d} + \eta_{\rm pen},$$
(1)

where  $v_{\rm w}$  is the welding speed, m/s;  $F_{\rm weld}$  is the area of weld cross section, m<sup>2</sup>;  $F_{\rm d}$  is the area of deposited metal section, m<sup>2</sup>;  $F_{\rm pen}$  is the area of penetration of base metal, m<sup>2</sup>;  $\gamma_{\rm m}$  is the specific density of metal, kg/m<sup>3</sup>, which equals 7850 kg/m<sup>3</sup> for low-carbon steel;  $H_{\rm melt}$  is the enthalpy at melting temperature taking into account open melting heat, J/kg,  $H_{\rm melt} = 1340$  J/g [5] is taken for low-carbon steel; Q = IU is the heat power of welding heat source, J/s;  $\eta_{\rm d}$ ,  $\eta_{\rm pen}$  is the total heat efficiency of process of deposition and penetration of base metal [9].

**Procedure of experiment.** Present paper studies the effect of technological parameters (bevel angle  $\alpha$ , position of electrode in bevel – distance between electrode tip and edge – coordinate x, as well as welding speed  $v_w$ ) on bead section area and total heat efficiency of process in mechanized gas-shielded welding for evaluation of dependence of wall bead formation on geometry of bevel and position of electrode in terms of formation of high quality wall bead.

Welding was carried out by bead passing in accordance with scheme, given in Figure 1. Samples from 09G2S steel (0.5-0.8 Si, 1.3-1.7 Mn, 0.30 Cu, 0.30 Ni, 0.040 S, 0.12 C, 0.035 P, 0.30 Cr, 0.008 N, 0.08 As) that are  $200 \times 500 \times 20$  mm plates simulating welded joint with 15, 25, 35° grooving and shoulder 5 mm similar to performed root pass were used. System for experimental studies (Figure 2) includes table with current supply, autocarriage Noboruder NB-5H and welding machine S5 Pulse SHTORM-LORCH. Parameters of welding mode were registered with the help of devices installed on the control panel of welding machine.

SG2 welding wire 1.2 mm diameter (analog Sv-08G2S according to GOST 2246–70) and mixed shielding gases 82 % Ar and 18 % CO<sub>2</sub> according to TU 2114-004-00204760–99 were used for welding.

Values of factors were changed in accordance with plan of full-factor experiment (Table).



Figure 2. General view of welding installation



	Value			
Factor				
	-1	0	+1	
Bevel angel $\alpha$ , deg ( $x_1$ )	15	25	35	
Distance from axis to electrode $x$ , mm $(x_2)$	0	1.5/2/4	2.5/4/7	
Welding speed, m/h ( $x_3$ )	18	24	30	

Welding mode was selected so as to provide satisfactory weld formation:

welding current <i>I</i> <sub>w</sub> , A	$217 \pm 10$
wire feed rate $v_{wf}$ , m/min	6.3
arc voltage $U_a$ , $\overline{V}$	214 ± 1
consumption of shielding gas, 1/min	18
electrode diameter d, mm	1.2
electrode stickout, mm	20 ± 1

Using measurement of manufactured macrosections (Figure 3) of welded joints, bead section area was determined, and total heat efficiency coefficient was calculated on formula 1.

On the basis of experimental data, regressive equation was obtained:

$$F_V(\alpha, x, v_w) = 47.498 - 0.031\alpha + + 4.205x - 0.956v_w - 0.113\alpha x + 0.003\alpha v_w - (2) - 0.143xv_w + 0.004\alpha xv_w (mm^2),$$

as well as total neat efficiency of welding process:

$$\eta_V(\alpha, x, v_w) = 0.1253 - 0.0055\alpha + 0.018x + 0.0239v_w + 0.0006\alpha^2 + 0.0006x^2 - (3) - 0.0005v_w^2 - 0.0005\alpha x + 0.00012\alpha v_w - 0.00013xv_w.$$

Analysis of results and their generalization. Analysis of obtained dependencies of bead section area and total heat efficiency on welding speed showed that they have complex nature during electrode movement within groove width. Namely under the same conditions  $(v_w, \alpha)$  during electrode movement to the edge the values of bead section area and efficiency increase, and under other conditions they decrease. Such dependence might be explained by mutual influence of groove geometry and parameters of welding mode, that characterize electrode (arc) position towards the molten metal interlayer. With increase of welding speed, bead section area reduces and total heat efficiency of welding process increases only to specific value. This might be explained by the fact that with increase of welding speed, quantity of metal being deposited per unit of weld length reduces [9]. However, at that, arc column starts to deviate sideway opposite to the welding direction with the increase of welding speed. Deviating arc column displaces part of liquid metal in tail part of the pool. Thinning of liquid interlayer under the arc provides the increase of penetration depth at rise of welding speed up to specific value. Depth of penetration decreases at further speed increase due to reduction of heat input.

Given results match well with data of papers [10, 11] from which we know that the speed of movement in liquid increases as it leaks to the bottom of crater and thickness of film at first rises, and then reduces. Thickness of film and, in particular, speed of metal movement in it significantly depend on parameters of welding (deposition) mode.

Besides, it is known that [12] distribution of specific power of heat flow over the surface of groove and weld pool has complex nature due to interaction of the arc with surface of the weld pool. Distribution of heat flow also significantly changes at alteration of welding mode due to change of form of the weld pool surface, as well as positioning of electrode in the grooves. Therefore, the most broad picture of heat influence of the arc in gap welding can be determined only using simulation of weld pool formation and ex-



Figure 3. Examples of macrosections of samples welded in downhand position



**Figure 4.** Diagram of scattering of experimental (*circles*) and calculated (*solid line*) values of area of bead section (*a*) and heat efficiency (*b*) of welding process

perimental studies taking into account the groove shape and specific welding mode.

Obvious conclusion about combined influence of parameters of welding mode and liquid interlayer under the arc on weld formation can be made based on the mentioned above. Therefore, it should be considered in the equations used for determination of bead section area and heat efficiency. In this case, they are represented in multiplicative form and written in the following way:

$$F = F_V(\alpha, x, v_w)\theta_F(I_w), \qquad (4)$$

$$\eta_{\rm w} = \eta_V(\alpha, x, v_{\rm w})\theta_{\rm n}(I_{\rm w}), \qquad (5)$$

where  $F_V(\alpha, x, v_w)$ ,  $\eta_V(\alpha, x, v_w)$  are the functions of dependence of area of bead section and heat effect on bevel angle  $\alpha$ , electrode position in groove x, speed of welding  $v_w$ ;  $\theta_F(I_w)$ ,  $\theta_\eta(I_w)$ are the functions of dependence of area of bead section and heat efficiency on welding current  $I_w$ , respectively.

Functions  $\theta_F(I_w)$ ,  $\theta_\eta(I_w)$  were determined in a course of experiment:

$$\theta_F(I_{\rm w}) = 0.0134I_{\rm w} - 1.559,\tag{6}$$

 $\theta_{\rm n}(I_{\rm w}) = 0.0047I_{\rm w} + 0.084. \tag{7}$ 

Conformity of obtained equations (4) and (5) were evaluated using diagrams of scattering of experimental and calculated values of bead section area and heat efficiency of the welding process (Figure 4).

Diagrams of scattering, given in Figure 4, show satisfactory matching of theoretical and experimental values of bead section area and heat efficiency of the welding process. Conformity control of obtained equations using Fisher *F*-criterion gave positive results that characterize their accuracy.

Problem (as inverse) on determination of modes of welding with set bead section area and coefficient of areas k was solved based on experimental and theoretical data. Given dependencies were determined and obtained in the following form:

• welding current

$$I_{\rm w} = \frac{-b + \sqrt{b^2 - 4ac}}{2a};$$
 (8)

coefficients *a*, *b*, *c* are determined using following formulae:

$$a = -0.015 \frac{\pi d^2}{F_{\rm d}}; \quad b = 0.637 + \frac{\pi d^2}{F_{\rm d}} (3.341 - 0.01\alpha);$$
  
$$c = -74.048 + 0.048\alpha - \frac{\pi d^2}{F_{\rm d}} (185.97 - 0.583\alpha) - (1 + k_{\rm f})F_{\rm d},$$

where  $k_{\rm f}$  is the coefficient of fusion in consumable electrode welding, it equals 0.18–4 at applied modes;

voltage

$$U_{\rm ef} = \frac{B_{\rm ef} v_{\rm w.f}}{\eta_{\rm w} I_{\rm w}};\tag{9}$$

coefficient  $B_{\rm ef}$  is determined on formula

$$B_{\rm ef} = \gamma_{\rm m} H_{\rm melt} \frac{1+k_{\rm f}}{4} \pi d^2;$$

• welding speed

$$v_{\rm w} = \frac{\pi d^2 (0.08 I_{\rm w} - 8.32) \cdot 60}{4F_{\rm d}} \,\,({\rm m/h}). \tag{10}$$

Experiments for verification of convergence of obtained dependencies in this study and of evaluation convergence of results of welding parameters determination using obtained equations (8)-(10) were carried out, and graphs of welding current values convergence (Figure 5) and diagram of scattering of actual and calculated voltage values (Figure 6) were built on their basis.

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**Figure 5.** Diagram of convergence of experimental (1) and calculated (2) values of welding current (error  $\pm 5$  %)

As can be seen from Figures 5 and 6, scattering of values does not exceed 10–12 %, therefore, these equations for determination of parameters of mode of wall bead deposition (also can be used for calculation of parameters of mode of deposition on inclined surface) can be used in development of technology of mechanized multi-pass shielded-gas welding.

#### Conclusions

1. General equations of dependence of welding process heat efficiency and bead section area on technological parameters of welding were formulated and obtained in analytical form. The equations (except for known parameters) take into account mutual effect of electrode position relatively to liquid metal pool ( $F_V(\alpha, x, v_w)$ ,  $\eta_V(\alpha, x, v_w)$ ) and value of interlayer of molten metal under the arc ( $\theta_F(I_w)$  and  $\theta_\eta(I_w)$ ). 2. Obtained equations allow calculating pa-

2. Obtained equations allow calculating parameters of deposition of first bead in the layer, which provide for set coefficient of areas  $k = F_{\text{pen}}/F_{\text{d}}$  (in considered limits of *k* from 0.20 up to 1.57), and as a result reduction of possibility of appearance of such defects as lack of penetration.

3. Optimum values of bevel angle  $\alpha = 20.3^{\circ}$ and speed of welding  $v_{\rm w} = 26.34$  m/h were determined for given parameters of mode with the help of obtained equations under condition of maximum value of total heat efficiency of welding process  $\eta_{\rm w} = 0.392$ . The most efficient heat input is provided at given values of bevel angel and speed of welding.

4. Program for calculation of wall bead deposition modes was developed based on results of performed work.



Figure 6. Diagram of scattering of experimental (*circles*) and calculated (*solid line*) values of voltage

5. Obtained analytical dependencies can be used in the future for development of algorithms to control of welding mode parameters in automatic and robotized units.

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