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STRESS-STRAIN STATE OF DIFFUSION BONDS BETWEEN METALS WITH DIFFERENT PHYSICAL-MECHANICAL PROPERTIES

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Regularities were established in the effect exerted by plastic strains on formation of the stress-strain state in diffusion bonding of pieces of dissimilar materials in the form of cylinders and sleeves under compression and variable temperature conditions. It is shown that the lowest level of plastic strains and their unfavourable distributions along the axis of cylinders and sleeves take place in bonding of materials under compression and constant temperature conditions.

Keywords: diffusion bonding, dissimilar materials, stressstrain state, plastic strains, compression, temperature

As shown by our earlier investigations of the fields of stresses within the elasticity ranges in vacuum diffusion bonding (VDB), the distribution of stresses and strains is far from being uniform, particularly in the connections made between dissimilar materials [1]. The volumetric stressed state caused by differences in rigidity and linear temperature expansion coefficients (LTEC) of the materials joined is formed within the bond zone. The character of the distribution of equivalent stresses in height of a specimen during thermal cycling leads to localisation of plastic strains particularly within the bond zone, while the highest equivalent stresses and expected plastic strains are observed on the bond periphery.

In real materials, VDB may also induce inelastic strains, which can be subdivided into the plastic ones that are independent of time (instantaneous plasticity ε_{ij}^p), and those that depend upon the time of diffusion plasticity (creep ε_{ij}^c). Strain tensor is a sum of $\varepsilon_{ij} = \varepsilon_{ij}^e + \varepsilon_{ij}^c + \varepsilon_{ij}^p$, where ε_{ij}^e are the elastic strains [2]. Formation of plastic strains is related to a certain level of the stressed state characterised by invariant σ_i (equivalent stresses or stress intensity). At $\sigma_i < \sigma_y$, where σ_y is the yield strength of a material, plastic strains of instantaneous plasticity, ε_{ij}^p , do not develop.

Plastic strains formed in bonding may affect the character of the field of stresses and strains, thus making it topical to investigate the effect of plastic strains on the stress-strain state (SSS) in VDB.

The purpose of this study was to establish regularities in formation of SSS in VDB of connections between dissimilar materials, having the cylindrical form of the type of cylinder--cylinder (C--C) and sleeve--sleeve (S--S), allowing for plastic strains formed in the connections.

At the present stage, we considered only the instantaneous plasticity strains. Investigations were conducted by the method of computer modelling using software package ANSYS on connections of the C--C and S--S types. Thermal-physical properties of the materials joined were assumed to be similar to those considered in the elastic problems [1], except for the yield strength, which in all the cases was assumed to be higher than minimal equivalent stresses and lower than the maximal ones obtained with the elastic solution ($\sigma_{min} < \sigma_y < \sigma_{max}$). Mises condition $\sigma'_{eq} = \sigma_y$, where σ'_{eq} , are the equivalent (reduced) stresses, was taken as a condition of formation of plastic strains.

For the axisymmetric problem (cylindrical structures), tangential stresses $\tau_{yz} = 0$ and $\tau_{xz} = 0$. Therefore,

$$\sigma_{\rm eq}' = \sqrt{\sigma_x^2 + \sigma_y^2 + \sigma_z^2 - \sigma_x \sigma_y - \sigma_y \sigma_z - \sigma_z \sigma_x + 3\tau_{xy}^2},$$

where σ_x are the radial stresses; σ_y are the axial stresses; σ_z are the circumferential (tangential) stresses; and τ_{xy} are the tangential stresses.

Investigations of SSS were conducted on the C--C and S--S connections, which are similar in shape and size to those considered in the elastic problem (Figure 1). Final element models of the connections are given in study [1]. The elasticity modulus was assumed to be identical for materials 1 and 2 to be joined ($E_1 = E_2 = 1.10^5$ MPa).

LTECs of the materials joined were assumed in all the variants, except for calculation model 1, to be two times different ($\alpha_1 = 1 \cdot 10^{-5} 1/\text{deg}$, and $\alpha_2 = 2 \cdot 10^{-5} 1/\text{deg}$). In model 1, consideration was given to loading only by an external force at a constant temperature. Therefore, LTECs of the materials joined were assumed to be identical ($\alpha_1 = \alpha_2 = 1 \cdot 10^{-5} 1/\text{deg}$). The investigated variants of combinations of materials are given in the Table.

Loading was varied as follows: only compression force (model 1), only heating or cooling (models 2 and 3), and combination of compression with heating or cooling (models 4 and 5). Yield strength of the materials joined was selected to be at such a level that plastic strains took place in both of the pieces joined, i.e. identical yield strengths (models 2 and 4), or only

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Variant	LTEC $\cdot 10^6$ of material, $1/\deg$		Loading	T, °C	σ _y of m M	naterial, Pa
NO.	1	2	<i>p</i> , Mra		1	2
1 _p	10	10	40	0	39	80
$2^{\rm h}_{ m p}$	10	20	0	+100	60	60
$2_{\rm p}^{\rm c}$	10	20	0	-100	60	60
$3^{\rm h}_{ m p}$	10	20	0	+100	60	120
$3_{\rm p}^{\rm c}$	10	20	0	-100	60	120
4_p^h	10	20	40	+100	80	80
4_p^c	10	20	40	-100	80	80
$5^{\rm h}_{ m p}$	10	20	40	+100	80	160
$5^{\rm c}_{ m p}$	10	20	40	-100	80	160
Note. I	ndices «h»	> and «c»	stand for	r heating a	nd cooli	ng, and

Investigated variants of combinations of properties of materials 1 and 2 in connections C--C and S--S (calculation models 1-5)

in one upper piece, i.e. $\sigma_{y, 1} = 0.5\sigma_{y, 2} < \sigma_{eq}^{max}$ (models 1, 3 and 5). Strengthening modulus in plastic deformation for all the materials was assumed to be equal to zero, except for variant 1, where it was assumed

index «p» -- plastic solution

to be equal to 1.10^3 MPa to provide consistency of the solution. In variant 1, consideration was given to SSS in VDB of materials with different yield strengths, which is characteristic of joining dissimilar materials, including brittle materials (ceramics, graphite) to ductile ones (metals). Results of the plastic solution were compared with those derived in a similar elastic problem, where equivalent stresses in both materials do not reach the yield strength value. Diagrams of the distribution of axial stresses, which provide drawing together of the mating surfaces in sections 1--1 along the bond and in section 3--3 located across the bond at a distance equal to half the radius of a cylinder in model 1, in VDB of cylinders are shown in Figure 2. The distribution of axial stresses within the bond zone

on connections S--S is the same. Analysis of the fields of stresses shows that in compression of materials with identical elasticity modulus E in connections of both types the uniform linear stressed state persists as long as both materials behave elastically. Appearance of plastic strains causes disturbance of the uniformity of the fields of stresses in a weaker material, and their symmetry about the bond is violated as well. In this case, equivalent stresses near the bond in a stronger material grow to some extent, compared with the rated ones (applied to a specimen). In a specimen of a weaker material, the similar zone of increased stresses is located far from the bond. The character of the fields of equivalent stresses in this variant indicates that the difference in strength of the materials joined in the case of the plastic solution shows up in a way similar to the difference in rigidity in the case of the elastic solution.



Figure 1. Schematics of specimens of bonds of the C–C (a) and S–S (b) types

The fields of plastic strains in VDB of materials with different yield strengths in cylinders and sleeves are indicative of the fact that equivalent stresses decrease in a material with a lower yield strength at the moment of appearance of plastic strains in it. Axial compressive stresses change in a similar way (Figure 2, b). As a result, further loading causes deformation of part of a piece located far from the bond.

Plastic strains form in a weaker metal, but near the bond they are low. Axial shortening is not in excess of 0.005 %, and plastic elongation in the radial and circumferential directions is still lower.

The fields of axial, radial, circumferential, tangential, principal and equivalent stresses, as well as the fields of plastic strains were studied for all the models and variants.

Diagrams of plastic strains (Figure 3) prove that in the case of compression and constant temperature in materials with a different strength, the plastic



Figure 2. Diagrams of axial stresses in sections 1--1 (*a*) and 3--3 (*b*) of C--C connections obtained with elastic solution (1_e) and with allowance for plasticity (1_p)



Figure 3. Diagrams of plastic strains (according to Mises) in sections 1-1 (a, b), 2-2 (c, d) and 3-3 (e, f) in C-C (a, c, e) and S-S (b, d, f) connections of models 1-5

strains develop in a weaker material outside the bond, this leading to increase in general strains and a nonguaranteed quality of the bond.

In model 2, SSS is considered under loading by heating and cooling of the materials with identical strengths and different LTECs without compression. As shown by analysis of the fields of stresses, in this case a complex field of stresses, symmetrical about the bond, is formed within the bond zone. The character of this field remains almost the same both in the area of elastic deformation and with appearance of plastic strains. The difference can be seen only within a very narrow (about 1 mm) zone located near the bond.

Analysis of the field of plastic strains in heating or cooling of the materials with different LTECs but identical yield strengths shows that this variant provides an ideal localisation of plastic strains particularly within the bond zone, covering both of the materials joined. Plastic strains lead to decrease and lev-

elling of equivalent stresses along the bond in both materials in heating and cooling. The level of stresses decreases. The distribution of equivalent stresses along the axis of a specimen preserves its ideal character in terms of localisation of plastic strains. However, the peak within the bond zone becomes blunt.

The distribution of axial and principal stresses along the bond is similar to purely elastic loading. In this case, the values of axial stresses in a major part of the bond are close to zero and dramatically grow only in a narrow zone near the bond edge. They change in character along the axis of a specimen and become close to sinusoid with maxima at a distance of about 0.2–2.5 mm from the bond.

As seen from Figure 3, plastic strains form in both materials in the immediate proximity to the bond, but their distribution along the bond is non-uniform. They are low in the middle part of the bond (about 0.01 %) and gradually grow (to 0.16 %) with distance to the external surface of a specimen (Figure 3, a-d). The point of minimum in the diagram of plastic strains is located in a cylinder on its axis, and in a sleeve --- at a distance of 0.23 of its thickness from its internal surface, i.e. at a point with zero tangential and minimal equivalent stresses in the case of the elastic solution. In the theory of pressure treatment of metals, the zone where the values of tangential stresses fall and change their sign is called the stagnant or stick zone (region). In study [3], such a zone is situated on the contact surface at the centre of a cylinder during its upsetting. Modelling shows the presence of the stagnant zone in VDB of not only cylinders but also of sleeves.

With distance from the bond, the values of plastic strains dramatically decrease (Figure 3, e, f). In a direction along the specimen axis, the values of plastic strains (shortening in a material with lower LTEC, and elongation in a material with higher LTEC) in heating are markedly in excess of those in the radial and circumferential directions, having the opposite signs. In cooling, the character of plastic strains persists, but shortening strains change into the elongation ones, and vice versa.

Therefore, this confirms the conclusion made with the elastic solution of the problem that thermal cycling in joining of materials with different LTECs creates ideal conditions for localisation of plastic strains particularly within the bond zone. Plastic strains develop mainly along the specimen axis, i.e. normal to the bond, which should promote physical contact of materials in the bond and improvement of its quality. However, non-uniformity of the distribution of plastic strains along the bond persists.

In model 3, the materials joined differ not only in LTEC but also in yield strength, i.e. plastic strains develop mostly in one of the materials. A complex field of equivalent stresses persists within the bond region in this model. The character of this field remains almost identical both in a region of elastic deformation and with formation of plastic strains in bonding the materials of the same and different strength, the dif-

ference can be seen only within a very narrow (about 1 mm) zone near the bond. In particular, the symmetry about the bond is violated. The distribution of axial and principal stresses along the bond and the character of the stressed state are similar to purely elastic loading and plastic deformation of both materials (model 2). Analysis of the fields of plastic strains in model 3 also showed their localisation within the bond zone, but only on the side of a material with lower strength. Plastic strains lead to decrease and levelling of equivalent stresses along the bond first of all particularly in a weaker material both in heating and in cooling. The level of stresses decreases, and their distribution becomes more uniform, compared with the results obtained in solutions within the elasticity range. The distribution of equivalent stresses along the specimen axis retains its ideal character in terms of localisation of plastic strains, but the peak within the zone of formation of plastic strains becomes a bit blunt.

Plastic strains non-uniformly distribute along the bond and concentrate in a material with lower strength (see Figure 3). They are low (about 0.01 %) in the middle part of the bond (at the stagnation point). However, they gradually grow to 0.32–0.34 % with distance to the external surface of a specimen (see Figure 3, a, b). With distance from the bond, the values of plastic strains rapidly decrease (see Figure 3, e, f). No plastic strains are present in a stronger material (see Figure 3, c, d).

The main regularities established for loading by thermal cycling in a bond between materials with different LTECs are also true for bonding of materials of different strength.

Model 4 considers SSS of materials with different LTECs and identical yield strengths, subjected to a combined loading by compression and heating–cooling.

The fields of equivalent stresses within the bond region become more complex with a clearly defined asymmetry. The presence of plastic strains changes the character of the field of stresses only in the immediate proximity to the bond (within a zone about 1 mm wide). When heating changes into cooling, the fields in materials 1 and 2 change places both at the absence of plastic strains and in plastic deformation of both materials.

The peculiarities and character of the fields of axial and principal stresses noted with the elastic solution persist. However, their values change to some extent.

Appearance of plastic strains is seen in turn in a material with lower LTEC in heating, and with higher LTEC in cooling, this leading to decrease and levelling of equivalent stresses along the bond in a plastically deformed material. The level of stresses decreases to some degree. In a material where plastic strains do not form (with higher LTEC in heating, and with lower LTEC in cooling), the non-uniformity of the distribution of stresses persists, but their level decreases. Along the specimen axis the distribution of equivalent stresses retains the character close to ideal in terms of localisation of plastic strains, but the peak



within the bond zone becomes somewhat blunt. In addition, it alternately covers one or the other material when heating changes into cooling.

As to the distribution of axial stresses, in this case the regularities are the same as those established for heating-cooling without compression. The average level of axial stresses shifts from zero to a level of applied axial compression (40 MPa). The character of the distribution of principal stresses along the bond with formation of plastic strains changes but slightly. And their level also changes insignificantly.

Plastic strains non-uniformly distribute both along the bond and between materials with different LTECs. In a material with lower LTEC in heating (with higher LTEC in cooling), within the stagnant zone they are at a level of 0.055 %, but with distance to the external surface of a specimen they gradually increase to 0.12 % (see Figure 3, a, b). In a material with higher LTEC in heating (lower LTEC in cooling), plastic strains are absent along the entire length of the bond (see Figure 3, *c*, *d*). The general level of plastic strains in combined loading by compression and temperature change is higher as a whole than in temperature change without compression (models 2 and 3). With distance from the bond, the values of plastic strains rapidly decrease (see Figure 3, e, f).

In heating, plastic strains of the shortening type form in a material with both lower and higher LTEC, and their values exceed markedly the values of deformation in the radial and circumferential directions, having opposite signs. In cooling, the character of the distribution of deformation persists.

Therefore, this confirms a conclusion made with the elastic solution of the problem that thermal cycling in bonding of materials with different LTECs creates ideal conditions for localisation of plastic strains particularly within the bond zone, as well as in a case of combined loading by compression and heating--cooling. Non-uniformity of the distribution of plastic strains decreases along the bond.

Model 5 considered SSS in compression and heating--cooling of materials with different values of strength and LTEC. It was established that within the bond region the fields of equivalent stresses retain a very complicated character. The presence of plastic strains changes the character of the field of stresses in the immediate proximity to the bond. Deformed in heating is only the material having a lower yield strength value, and only in the case if it has lower LTEC. In cooling, only elastic deformation of both materials takes place.

In the case of compression with heating, equivalent stresses in a material with lower strength and lower LTEC decrease and uniformly distribute along the entire bond. In a stronger material, equivalent stresses also decrease to some extent, but non-uniformity of their distribution along the bond persists.

The distribution of equivalent stresses along the specimen axis has a character close to ideal in terms of localisation of plastic strains, but the peak within the bond zone becomes blunt in heating on the side of a weaker material.

Plastic strains of the shortening type form in heating along the specimen axis. Their values are markedly higher than those in the radial and circumferential directions, having opposite signs.

In compression with cooling, plastic strains are absent in the materials having either higher or lower strength. It is necessary to increase the compression pressure or the range of temperature changes for the plastic deformation to occur.

Therefore, as shown by analysis of the fields and diagrams of plastic stresses, VDB of materials with different physical-mechanical properties under compression and thermal cycling leads to formation of a complex stressed state, which localises plastic strains within the bond zone, thus intensifying the bonding process. And it is this fact that was used as a basis for the new VDB method [4].

Plastic strains reach the highest values (0.055 %at the stagnation point, and 0.120 % on the periphery) in model 4 (combined compression and thermal cycling with identical yield strengths and different LTECs). This also provides the most uniform distribution of plastic strains along the bond.

CONCLUSIONS

1. Analysis of available data showed that the regularities established earlier within the elasticity range persist as a whole, but the solution results obtained with allowance for plasticity provide a more complete and objective picture of SSS in VDB of dissimilar materials. In particular, only the solution derived with allowance for plasticity makes it possible to estimate the degree of uniformity of the distribution of plastic strains along the bond.

2. The most uniform distribution of plastic strains in the bond is provided by compression combined with thermal cycling on materials with close values of yield strength. Plastic strains form in turn in a material with lower LTEC in heating and higher LTEC in cooling.

3. The most uniform distribution of plastic strains in the bond takes place in thermal cycling of materials with different strength values without subjecting them to compression.

4. In compression with thermal cycling of materials with different strength values, plastic deformation occurs only at a stage of heating. The strains are absent at a stage of cooling.

5. The lowest level of plastic strains and their unfavourable distribution along the axis of cylinders and sleeves take place in VDB of materials having different strengths under the conditions of compression at a constant temperature.

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IMPROVEMENT OF ChS-104 NICKEL-BASE ALLOY WELDABILITY BY OPTIMIZATION OF HEAT TREATMENT MODE

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The paper deals with weldability of a high-temperature nickel-base alloy with dispersion strengthening. It is shown that the alloy cracking resistance can be increased by optimization of the heat treatment mode before welding and lowering the cooling rate after welding.

Keywords: thermal hardening nickel alloys, γ -phase, arc welding, hot cracks, heat treatment, overageing, long-term strength

Cast nickel alloys with dispersion strengthening belong to un-weldable or poorly weldable structural materials as a result of γ -phase precipitation. The reason for that is increased sensitivity of welded joints to crack formation when welding, as well as when performing postweld heat treatment. The above-mentioned concerns fusion welding methods, for which concentrated heating that creates considerable temperature gradients in heat-affected zone (HAZ), is typical. The use of uniform heating for obtaining welded or brazed joints prevents crack formation, as a rule, though such methods of joining are characterized by considerable labour consumption and have limited capabilities.

It is known that deterioration of weldability and cracking resistance of nickel alloy welded joints at postweld heat treatment are in direct dependence on γ -phase content. That refers in full measure also to the ChS-104 casting alloy, used at the enterprise for manufacturing of nozzle blades of all the stage in turbine part of gas turbine installations that operate at the temperature up to 950 °C. Chemical composition of this alloy is the following, wt.%: 0.07--0.14 C; 20.0--21.8 Cr; 10.3--12.0 Co; 0.3--0.9 Mo; 3--4 W; 2.1--2.9 Al; 3.1--3.9 Ti; 0.15--0.35 Nb; ≤ 0.3 Mn; ≤ 0.3 Si; ≤ 0.008 S; 0.008 P; 0.015 B; ≤ 0.5 Fe; Ni ---base. It is obvious from the given chemical alloy composition that the weight fraction of titanium and aluminium in it is enough for weldability deterioration. In accordance with theoretical evaluation of welded joint resistance to crack formation at postweld heattreatment, the alloy is in the most unfavorable area (Figure 1), and multi-step overageing is required before welding for prevention of crack formation [1].

The alloy is lends itself easily to brazing, although fusion welding is required for joining the blades into «packets» and restoration of their complicated geometrical sizes in case of casting rejects, namely electron beam welding for «packet» joining and argon-arc welding (AAW) for blades sizes restoration. In such a way repairing cast defects by brazing is performed in vacuum furnaces on those parts of the blade (on airfoil), where qualitative correction of the defect by welding is impossible. Where the correction by welding is possible (on band flanges) AAW with EP533 filler wire is used. It should be marked that the choice of this type of the wire is the best from the point of view of strength characteristics and corrosion resistance of welded joints [2–5].

Heat treatment recommended by the designers (R&D Institute of Structural Materials «Prometej») for the given alloy consists of the following operations: homogenizing at 1170 °C, 4.5 h, cooling in the air; ageing at 1050 °C, 4.5 h, cooling in the air; the same at 850 °C, 16–17 h, cooling in the air.

It is accepted that such order of operations with quick cooling (in air), and that is very important, allows getting the best strength properties of material.



Figure 1. Influence of titanium and aluminium content on nickel alloys sensitivity to crack formation in heat treatment of welded joints: *I* — homogeneous and semi-ageing alloys (γ -phase content of 3–5 vol.%) non sensitive to cracks formation in welding and heat treatment; *II* — precipitation-hardening alloys (γ -phase content is not more than 18–20 vol.%) with moderate sensitivity to crack formation; *III* — precipitation-hardening alloys (γ -phase content is more than 20–25 vol.%) with high sensitivity to crack formation

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Figure 2. Microstructure of ChS-104 alloy HAZ section (×100)

Moreover, in accordance with classical theory for crack formation prevention it is necessary to welding such alloys in the solid solution quenched state (after homogenizing with quick cooling) [5, 6]. However, when AAW is used with EP533, EI602, EP648 filler wires, formation of 0.2--3.5 mm long crack was observed in the ChS-104 HAZ alloy after every step of heat treatment (Figure 2). Checking was carried out by luminescence crack detection method and by metallographic examination. Decrease of heat input into the part did not give a positive effect ---- the cracks did not disappear. They were found directly after welding without further heat treatment. It should be marked that when welding in fully aged state the number of found cracks was 3.5 times less than when the alloy was welded after homogenizing (23 cracks in ten repaired places in comparison with 81 cracks at the same number of repairs).

It follows from the obtained results that if the cracks appear in HAZ (cracks do not form in weld

metal), then the base metal should be prepared for welding by the way of preliminary overageing [7--9].

The designer gives also the second mode of heat treatment that consists of homogenizing at 1170 °C, 4.5 h, cooling in the furnace up to 900–950 °C, soaking for 2 h and then cooling in air, ageing at 850 °C, 17 h, cooling in air.

Material weldability should be improved at such heat treatment, however short- and long-term strength decrease by 6--9 % (by 1.3--2.3 times) correspondingly depending on the temperature of tests. A number of blades and «packets» are subjected, according to the technology, to the accepted operation of alumosiliconizing that means carrying out diffusion annealing at 1030--1050 °C. In such a way the effect of ageing is removed and it is necessary to repeat this operation after welding. Weldability improvement after full heat treatment by the second version does not happen at quick cooling (from 1170 to 950 °C) during 15--20 min. It was found that the rate of cooling after homogenizining influences alloy weldability to the highest degree. Cooling rate should be 2-31 °C/min at temperature ranges of 1170-950 °C for prevention of cracks formation after welding. The following cooling rate does not influence the weldability. Heat treatment by the established technology (ageing at 1050 °C, 4 h, cooling in air and ageing at 850 °C, 16-17 h, cooling in air) does not lead to crack formation.

As it is known, coagulation of γ -phase happens at extended aging that has a negative effect on high-temperature strength properties of the material [7, 9]. This was checked in practice on the samples from the base material (unwelded) (Table).

The obtained results testify to little decrease of long-term strength with preservation of short-time

Influence of heat treatment mode on mechanical properties and long-term strength of ChS-140 alloy at T = 900 °C

		Tests				
Heat treatment mode		Short-t	erm		Long-term ($\sigma = 205$ MPa)	
	σ_t , MPa	σ _{0.2} , MPa	δ, %	ψ, %	<i>t</i> , h	
Mode 1 Homogenization (1170 °C, 4.5 h, cooling in air) Ageing (1050 °C, 4.5 h, cooling in air) Ageing (850 °C, 17 h, cooling in air)	505	450	20	27	104	
Mode 2 Homogenization (1170 °C, 4.5 h, cooling in the furnace during 2 h up to 950 °C, cooling in air) Ageing (1050 °C, 4.5 h, cooling in air) Ageing (850 °C, 17 h, cooling in air)	500	430	25	42	100	
Mode 3 Homogenization (1170 °C, 4.5 h, cooling in the furnace during 2 h up to 950 °C, cooling in air) Homogenization (1170 °C, 4.5 h, cooling in air) Ageing (1050 °C, 4.5 h, cooling in air) Ageing (850 °C, 17 h, cooling in air)	500	400	21	32	150	

Notes. 1. Plant technical documentations requirements to mechanical properties at short-term tests: $\sigma_t \ge 46$ MPa, $\delta \ge 7$ %, $\psi \ge 14$ % (samples by GOST 9456); time before fracture at long-term tests $t \ge 50$ h (samples by GOST 10145). 2. Average values of six measurements are given.





Figure 3. Microstructure of ChS-104 alloy with γ -phase precipitation, obtained after heat treatment by mode 1 (*a*), 2 (*b*, *c* — in axial and inter-axial areas correspondingly) and 3 (*d*) (×10000)

strength and higher short-term ductility at the second heat treatment version. However proceeding from the operation experience, the given fact is not critical and it is not expedient to perform the second homogenizing after welding with quick cooling for recovery of strength characteristics of the base metal. Carrying out secondary homogenizing of welded joints is dangerous as the probability of crack formation is high. At that the cracks are located across the weld and their length can reach 12 mm. It is possible to decrease the intensity of crack formation by the way of maximum increase of heating rate in the range of ageing temperatures.

Morphology of γ -phase, obtained on scanning electron microscope «Camebax», is given in Figures 3 and 4. γ -phase precipitations, formed at complete heat treatment by mode 1, are shown in Figure 3, *a*, at that its size and distribution are the same in the axial and inter-axial areas; γ -phase at complete heat treatment by mode 2 in the axial and inter-axial areas,

correspondingly, is shown in Figure 3, b and c; γ -phase at complete heat treatment by mode 3 is shown in Figure 3, d, at that its size and distribution in axial and inter-axial areas are the same. γ -phase morphology directly after homogenizing with delayed cooling in the furnace from 1170 to 950 °C during 2 h in axial and inter-axial areas is shown in Figure 4, a, b.

As it is seen from the given Figures, γ -phase has approximately the same size (basically 0.15--0.20 µm) after heat treatment by modes 2 and 3 and distribution in axial area. However, γ -phase in inter-axial area at mode 2 is substantially larger (0.3 µm) and is distributed less frequently. It is possible that γ -phase morphology in inter-axial area is exactly what has the biggest influence on weldability improvement at overageing. γ -phase is finer (0.05--0.10 µm) and distributed more uniformly at heat treatment by mode 1. γ -phase is the most coagulated directly after homogenizing with delayed cooling, its size is 0.25--0.30 µm; that leads to relaxation resistance decrease and duc-



Figure 4. Microstructure of ChS-104 alloy with γ -phase precipitation, obtained after homogenization with delayed cooling in the furnace (1170–950 °C) during 2 h in axial (*a*) and inter-axial (*b*) areas, as well HAZ metal at all heat treatment modes (*c*) (×10000)



tility increase of the material. More infrequent distribution of γ -phase (in comparison with axial area) and irregularity of its shape are also observed. γ -phase is completely dissolved in the matrix after homogenizing and cooling in air and is not discovered at magnification of $\times 10000$. γ -phase size and distribution are the same under all heat treatment modes (Figure 4, c) in the ChS-104 alloy HAZ directly near the fusion line.

It should be marked that melt purity acquires great importance in solving crack resistance problem. So, at postweld heat treatment by mode 2 cracks of 0.1--0.3 mm length were nevertheless found in one corrected place in the amount of 2--3 pieces at secondary remelting when using scrap up to 80 %. In the case of pure charge use the cracks were not observed, however, in practice this mode can seldom be realized. In operation generated microcracks start developing, and after 20--33 thousand hours in blade sections, where the repair was done with the help of AAW, appearance of considerable number of cracks is observed. Despite absence of cases of repaired sections tearing off in the process of gas turbine installation operation, or damage to air-gas channel, the probability of this should be eliminated.

CONCLUSIONS

1. Essential increase of cracking resistance was achieved for ChS-104 alloy in welding and after postweld heat treatment due to decrease of cooling rate to 2--3 °C/min in the temperature range of 1170--950 °C at homogenizing before welding.

2. It is necessary to design filler material with principally different system of alloying, as well as the process of producing welded joints with its use, for guaranteed absence of cracks in welding and after postweld heat treatment, as well as with the aim of providing maximum high level of heat resistance of the base metal and balanced life of welded joint metal.

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APPLICATION OF LASER TECHNOLOGY FOR SINTERING OF THE TOOL COMPOSITES CONTAINING DIAMONDS

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The analysis of problems and results of researches in the field of non-isothermal sintering methods of the diamonds containing composites are presented in this paper. Results of experimental study of the influence of a laser irradiation on qualitative characteristics of diamonds, on physical and chemical processes on the diamond-sheaf border are submitted; the data indicating the existence of certain set of the regimes of laser irradiation that preserve the initial durability of diamonds, and some certain regimes that lead to the crack formation, oxidation and other negative processes are presented. Some problems of the development of the technology of coupled thermal and deformational laser sintering of composites, means of its realization, results of theoretical and experimental studying of the thermal processes that occur at liquid-phase sintering, basic technological parameters are discussed in detail. Results of practical realization of the developed process are shown.

Keywords: laser sintering, composites, diamonds, binders, tools, modeling, durability

The trend in modern mechanical machining is using universal tools to produce components of different shape and size with high durability and reliability and an ability to perform the whole cycle of material processing with just a one tool. These tools must have a reasonably high cutting ability and long lifecycle. Moreover, they must be of a very complicated shape (sphere, thorus, ellipsoid, hyperboloid, etc.) with a specific geometry of cutting elements in order to meet the requirements of cutting kinematics. To meet these demands such tools should be made of special super hard materials.

The goal of such tools, including diamonds, is to obtain shape and size, maintain strong bonding of the diamonds and securing the cutting ability throughout there lifetime.

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Figure 1. Ways of toolmaking composites manufacturing

There are different ways to manufacture these tools (Figure 1) [1--3]. These technologies differ in shaping surface profiles, diamond attachment, types of binders, etc. They have different advantages and disadvantages as well as areas of useful application.

Havig analized the manufacturing technologies for such tools it is possible to conclude that all of them have relatively low productivity. The market demand significantly exceed the production rate of catting tools and diamond grinding wheels in particular. There is a limited amount of bonding materials with relatively low melting temperatures (no higher than 700 °C). This fact significantly limits its optimal use for the tools which are used for cutting of the workpieces with variable mechanical properties (hardness, chemical composition etc.). Moreover, these bonding materials do not hold the diamond grains firmly thus limiting the cutting productivity, tool resistance, diamond consumption and cutting costs. This problem deepens with the decrease of diamond layer thickness with its strength reduces significantly. The existing manufacturing methods for diamond cutting tools can not guarantee the complete control over the location of diamonds, and, moreover, can not result in the manufacturing of multy-layer cutting tools.

The most efficient way of manufacturing is layerby-layer laser sintering of powder composites.

As a heating source the laser irradiation has various advantages. Due to the non-contact energy introduction into the material and its precise dosing it is possible to conduct a super-heating of local areas of the material in a broad range of temperatures. This allows using a broad spectrum of binders, including those, which require heating temperatures corresponding to the oxidation temperature of diamond and to secure the metallurgical bonding between diamond containing layer and steel body of the tool, increasing its strength. The locality of laser irradiation enables the formation of a single- and multi- row working elements of tools by a layer-by-layer method. Moreover, it is easy to control the concentration of components in the composite, to harden the binder and to obtain highly disperse structures.

The possibilities of the application of laser irradiation for the sintering of the materials is studied in [1, 2, 4, 5]. In [4, 5] laser sintering of powders from nickel, molybdenum, oxides, carbides, nitrides and borides on a steel substrate was done experimentally. Powder sintering in solid phase [4] produced no positive results although solid phase laser sintering of ultra-dispersed nanoscale powders was theoretical pronounced [1]. It was concluded that two processes of liquid-phase sintering are the most promising: laser sintering with low-melt binders and peripherial melting of separate powder particles [1, 5–8].

This paper deals with the influence of laser irradiation on the properties of diamond-containing tool composites and with its liquid-phase super-speed heating by means of laser sintering process.

Modeling of laser sintering process. Structural and phase changes in the materials are determined by the temperature regime of laser heating and further material cooling. The temperature field, heating and cooling rates depend on a number of factors, connected with the heating source, the material and the interaction process. The links between major factors and parameters of laser sintering process are shown in Figure 2.

In order to conduct the experimental investigations of laser sintering of composites and super-hard materials we need to obtain the necessary data that would describe the interconnections between the system «matrix--steel tool case--diamond containing composite» and main technological parameters: beam power P(W) and its distribution along the irradiate surface P(x, y) (W/cm²), focal spot diameter $d_0 = 2r_0$ (mm), beam travelling speed v (m/min), frequency f (Hz) and scanning amplitude B (mm).

Physical model of such a system is shown in Figure 3.





Figure 2. Links between major factors and parameters of laser sintering process of a tool composite

Therefore a non-steady non-linear uniform equation of heat transfer in the Cartesian coordinate system $X(x, y, z) \in \mathbb{R}^3$ was used for the analysis of temperature distribution:

$$c_{pi}(t)\rho_{i}(t)\frac{\partial t}{\partial \tau} = \frac{\partial}{\partial x}\left(\lambda_{i}(t)\frac{\partial t}{\partial x}\right) + \frac{\partial}{\partial y}\left(\lambda_{i}(t)\frac{\partial t}{\partial y}\right) + \frac{\partial}{\partial z}\left(\lambda_{i}(t)\frac{\partial t}{\partial z}\right), \quad \tau > 0; \quad i = \overline{1.3},$$

where c_{pi} is the heat capacity; ρ_i is the density; ρ_i is the heat conduction coefficient; *t* is the temperature, τ is the time.

The general physical model looks like: for the binder (i = 1):

$$\begin{cases} t < t_m - \frac{\Delta t}{2}, \ \lambda_1(t) = \lambda_s, \ [c_p(t)\rho(t)]_1 = c_{ps}\rho_s; \\ t_m - \frac{\Delta t}{2} \le t \le t_m + \frac{\Delta t}{2}, \ \lambda_1(t) = \lambda_s + \frac{\lambda_m - \lambda_s}{\Delta t} \ (t - t_m + \frac{\Delta t}{2}); \\ [c_p(t)\rho(t)] = c_{ps}\rho_s + \frac{c_{pm}\rho_m - c_{ps}\rho_s}{\Delta t} \ (t - t_m + \frac{\Delta t}{2}) + \frac{L_f}{\Delta t}; \\ t > t_m + \frac{\Delta t}{2}, \ \lambda_1(t) = \lambda_m; \ [c_p(t) \ \rho(t)]_1 = c_{pm}\rho_m; \end{cases}$$

for the tool case (i = 2):

$$\lambda_2(t) = \lambda_2; \ [c_p(t)\rho(t)]_2 = c_{p2}\rho_{p2};$$

for the matrix (i = 3):

$$\lambda_3(t) = \lambda_3; \ [c_p(t)\rho(t)]_3 = c_{p3}\rho_{p3},$$

where t_m is the melting point; Δt is the interpolation interval; λs , c_{ps} , ρ_s are the thermal conductivity, specific heat and binder density in solid state; λ_m , c_{pm} , ρ_m are the thermal conductivity, specific heat and molten binder density; L_f is the latent heat of binder 1st kind phase transition; λ_2 , c_{p2} , ρ_2 are the thermal conductivety, specific heat and density of the tool case; λ_3 , c_{p3} , ρ_3 are the thermal conductivity, specific heat and matrix density.

Initial conditions:

$$T(x, y, z)|_{\tau=0} = F(x, y, z) = t_{\text{medium}},$$

where t_{medium} is the ambient temperature;

C .

boundary conditions on outer surfaces at $\tau > 0$:

$$\Gamma_1: -\lambda \frac{\partial t}{\partial n} = q_r; \quad \Gamma_2: \frac{\partial t}{\partial n} = 0_r; \quad \Gamma_3: -\lambda \frac{\partial t}{\partial n} = \alpha(t - t_{\text{medium}});$$

boundary conditions in the contact zone Γ_4 at $\tau > 0$:

$$\begin{cases} t|_{\Gamma_4} = t|_{\Gamma_4^+}, \\ -\lambda_- \frac{\partial t}{\partial n}\Big|_{\Gamma_4^-} + \lambda_+ \frac{\partial e}{\partial m}\Big|_{\Gamma_4^+} \end{cases},$$

where *n* is the normal to the surface ; q_r is the power density; α is the heat transfer coefficient; Γ_1 is the irradiated surface; Γ_2 is the axial symmetry surface; Γ_3 are the surfaces contacting the outer medium; Γ_4 is

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Parameters	13Kh	45	12Kh18N10T
Heat conductivity, W/(cm·K)	26	30	27.8
Specific heat, J/(kg·K)	0.578	0.662	0.7
Melting temperature, °C	1535	1535	1400
Density $\cdot 10^3$ kg/m ³	7.77-7.85	7.777.85	7.98.2
Heat of phase transition ·10 ⁵ , J/kg	0.9	1.0	0.8

Table 1. Composite irradiation conditions

the boundary surface between the matrix and the binder.

The thermal process was simulated by ANSYS7. A parametric program was developed to conduct the computations, where all data needed for the modeling was considered: model size, thermal properties of the materials, heat-affected zone and laser interaction time. The conditions of irradiation are shown in Table 1.

The samples for modeling were disk body: steel 13Kh; matrix: steel 45; binder: based on steel 12Kh18N10T (Table 2).

The following assumptions were made:

• because of the high scanning speed and the small amplitude it was assumed that each point of a scanned surface (7 \pm 0.7 mm) faces the light source with a power density of 104 W /cm²;

• the same thermal properties were applied for the binder (12Kh18N10T) as for the monolithic material;

 \bullet the absorptivity of material was supposed to be constant.

A finite element method was implemented for the calculation of the temperature field distribution. The synthetic diamonds in composite was neglected. The computational domain was transformed into a grid of elementary tetrahedrons where each of them has a number corresponding to a material with definite thermal properties. Near the heat-affected zone the step of segmentation was decreased to get more precise calculation results.



Figure 3. Physical model and its boundaries

Experimental set-up and research methodology. A specially designed industrial laser system was used for laser sintering of composites and super-hard materials. This system consists of a high-power $\lambda = 10.6 \ \mu m \ CW \ CO_2$ laser, beam delivery system and programmable 5-axis coordinate table. The laser power in a range of P = 200--1500 W was focused with a $F = 300 \ mm$ lens. The spot size could be changed from $d_0 = 0.7$ --8.0 mm, and the speed v = 0.2--4.8 m/min.

A primary goal was to find the maximum temperatures at which no deterioration of strength is observed for different synthetic and natural diamonds.

This problem was solved in two ways: first by straight irradiation of diamond grains on a graphitic substrate (Figure 4, a) and second by indirect heating of diamonds with laser melting of surrounding and adjacent powder materials (Figure 4, b, c). After the laser treatment, the diamonds were tested for static strength using common techniques.

The study of the influence of laser irradiation on the nature of material structure (hardness, dispersive, uniformity, distribution of alloying elements) was analyzed on binders' of the following composites: PC 12NBK01-Co-diamond and XTN-Co-diamond, both of which have different basic elements. PC-12NBK01-Co-diamond is a material on nickel base, a mechanical mixture of powders: 65 % PG-10N-01 and 35 % WC. XTN ---- powder alloy on the basis of a stainless steel 12Kh18N10T with hardening phases TiB₂ and CrB₂.



Figure 4. Diamond irradiation schemes: direct irradiation (*a*), binder and diamonds melting at different times of influence (*b*), binder and diamonds melting with their simultaneous relatively to the laser beam (*c*): 1 - - laser beam; 2 - - lens; 3 - - diamonds; 4 - - diamonds in binder; 5 - - graphite substrate; 6 - - steel plate; 7 - - binder; 8 - - steel

Material	Fe	Ni	С	Cr	Ti	В	Al	Si
PG-10N-01	37	Base	0.61.0	14-20	-	2.8-4.2	0.81.2	4.0-4.5
XTN-23	Base	6-8		20.0-20.5	2.4-2.5	2.5-2.6		

Table 2. Chemical composition of powder materials

The chemical composition of PG-10N-01 and XTN-23 is listed in Table 2.

Samples of materials as packed beds of powder composites were treated in air with laser radiation at power densities from $1.4 \cdot 10^3$ to $1.4 \cdot 10^4$ W/cm² at speeds from 0.2 to 2.0 m/min. The samples have been analyzed by a raster electron microscopy and local X-ray spectral element analysis.

Experimental results. The synthetic diamonds of the types ACC125 (500/400), ACC125 (425/300) and ACC160 (400/315), taken in batches of 60 pcs selected by color (yellow and green-yellow), were subjected to the laser irradiation at conditions based on the previously calculations. The laser heating regimes were carefully calculated to refine the structure of sintered parts and to make them appropriate for a great deal of binding materials.

It is hard to forecast the absorbed energy at direct irradiation of diamonds (see Figure 4, a) because they are partially transparent for the laser light. Therefore, parallel grains of diamonds were subjected to indirect heated by laser melting of powders of Co, PC12NBK



Figure 5. Value of breaking load for the selected diamonds, original (*I*) and irradiated (*II*) at the regimes: $1 - W_p = 1.7 \cdot 10^3 \text{ W/cm}^2$, $\tau = 0.09 \text{ s}$; $2 - 0.6 \cdot 10^3 \text{ W/cm}^2$, 0.15 s; $3 - 0.4 \cdot 10^3 \text{ W/cm}^2$, 0.18 s; $4 - 0.2 \cdot 10^3 \text{ W/cm}^2$, 0.24 s (*a*), and optimal regimes for laser sintering of various types of binders (*b*)

and XTN-23, in which they have been placed previously (see Figure 4, b, c). The diamond strength test is presented in Figure 5, a.

The results demonstrate that laser irradiation can, to some extent, reduce the strength of diamond grains. First of all, this is caused by the appearance of cracks, which form on a surface of grains and are spread along their ribs. Cracks are caused because of the internal stresses of the diamonds, presence of defects, irradiation condition and most of all its duration. It may be concluded that defectless and strong synthetic diamonds at selected conditions of laser irradiation practically do not lose strength (see Figure 5, *a*). Figure 5, b shows the intervals of irradiation regimes $(W_{\rm p}, \tau)$ for different binding types where the initial quality of diamonds remains unchanged (Figure 6, a). Outside of these regions there is no sintering at all (left side) or the crack formation is observed (Figure 6, b) or the oxidation of diamonds occures (right side).

The analysis of these diamond grains has shown that there has been a good thermal contact to molten metal. This was proved by raster electron beam microscopy (Figure 7, a). On the surface of diamond grains at 180 Å up to 400 Å thick layer of molten cobalt was found using qualitative and quantitative microanalysis (Figure 7, b).

Since the melting point of cobalt is 1450 °C, it is reasonable to suggest that the diamond grains were heated to the same temperatures but no noticeable thermal destruction of diamonds was observed. The presence of a film of metal shows also that the diamonds excellent wetted by cobalt, which is crucial for the mechanical fixing of diamonds by the binder. It is also known that liquid cobalt dissolves carbon, which may have a negative effect on the mechanical properties of diamonds. However, at such small life times of the liquid phase $(10^{-2}-10^{-3} \text{ s})$ there was no noticeable diffusion. Nevertheless, to eliminate this effect it is recommended to use cobalt in binders together with carbon binders, for example tungsten carbide. At laser melting of powders XTN-23 and PC12NBK, with lower melting points (1220 and 1320 °C accordingly), neither thermal destruction of diamonds nor strength loss was found. Occasional initiation of cracks was observed, which may be considered as outcome of the initial presence of defects in the diamond grains. Previous sintering experiments of a system PC12NBKC1 (100 %) ---- diamond at power density of $5.1 \cdot 10^7$ --2.0 $\cdot 10^8$ W/cm², beam diameter of 3 to 5 mm at a scanning speed of 0.8 m/min did show cracking of the coating.

To increase the plasticity cobalt (90 %) was added to the powder PC12NBKC (10 %). With lower cobalt



Figure 6. Diamond sintered with optimal parameters lying within the confidence interval (*a*) and outside of the limits, with cracks ($\times 1000$) (*b*)

concentration it was practically impossible to get a satisfactory strength. The conditions of laser sintering were as the following: power density of $5.1 \cdot 10^7$ to $1.4 \cdot 10^8$ W/cm², beam diameter 3--5 mm and scan rate 0.8-1.2 m/min.

We may conclude that for a sintering of composition PC12NBKC1 (10 %) – Co (90 %)-diamond, the radiation power density should be $5 \cdot 10^7 \text{ W/cm}^2$, beam diameter 5--6 mm and scan rate 0.8–1.2 m/min. The conditions of sintering with power density of $1.4 \cdot 10^4 \text{ W/cm}^2$ are «critical», because here the cobalt is melting intensively.

Sintering composite of a system XTN (10 %)--cobalt (90 %)--diamond was performed at power density of $5.1 \cdot 10^3$ to $1.4 \cdot 10^4$ W/cm², beam diameter 3--8 mm, and scanning speed 0.8--1.2 m/min. At the upper limits ($W_p = 1.4 \cdot 10^4$ W/cm², $d_0 = 3$ mm, v == 0.8 m/min) the powder was melted. It was considerably porous at the diamond--binder interface. The structure of the molten material consists of cobalt grains surrounded with Co--Ni phase and impregnations of a Co--W carbide phase.

The grain size of cobalt varied from 5 up to $8 \mu m$. After laser processing the tungsten carbide content is less by factor 2.0–2.5 than in the initial powder. From our point of view the temperature in the irradiated zone considerably exceeded the cobalt melting point in this case. As a result, crack formation is initiated on the diamond surface. Some cracks of width not less than 1.3 μm metal drops were observed. A qualitative analysis of the droplet material showed that a ratio Fe–Co in the composite was 4.3 %, and in an ostium of a crack was 1.4 % respectively. This proves that in an ostium of a crack there are metals coming from the external surface after crack formation due to the influence of laser irradiation.

In general, the investigation results indicate that in this case, as well as with the sintering of composition PC12NBKC1 (10 %)--Co (90 %)--diamond, the beam power density should be within the limit of $5\cdot10^7$ W/m², the beam diameter should be within a range of 5--6 mm and scanning speed should not exceed the range of 0.8--1.2 m/min.

Laser irradiation allows sintering of all the composites given that there is no influence on the diamond monocrystals.

The metallurgical surveys have shown that after laser sintering the composite binder from alloy XTN has a highly dispersive structure (Figure 8, a). The existence of crystals of hardening phases (TiB₂, CrB₂) and eutectic are much higher than for plasma coatings. In molten layers the segments similar to martensite needles, are observed. They represent local eutectics with dispersed structure.

The hardening phases and γ -phase are oriented in the direction of heat removal. The cooling rate in near-surface layers was approximately $5 \cdot 10^4$ -- $10^5 \,^{\circ}C/s$. With increasing distance from the surface, the dispersivity of structure decreases and the dimensions of hardening phases increase. The microhardness of structure of melted layer changes smoothly with depth in the range 7000–7500 MPa depending on



Figure 7. Diamond grains coated with binder (a) and with film of cobalt (b)



Figure 8. Structure of the XTN-based composite after laser sintering, magnification (*a*) and hardness vs. depth of spray-coated XTN specimen at different working conditions: $t - W_p = 6 \text{ kW/cm}^2$, v = 0.2 m/min; $2 - 11 \text{ kW/cm}^2$; $3 - 25 \text{ kW/cm}^2$ (×400)

working conditions (Figure 8, b). The hardness of alloyed layers increases with the dispersion level of its components, the matrix hardness and with the increase of the number of harder eutectic phases.

The tribologic properties of composite binders obtained by laser melting (value and nature of wear, friction coefficient) were examined under conditions of dry friction. The binders from XTN and PC12NBKC1 were studied. The tests were conducted at pressures of 1 to 9 MPa. The slip velocity was 0.1 m/s. For comparison, the friction tests were performed for an XTN coating, obtained by plasma spraying.

The greatest wear resistance was found with XTN coating, having smaller hardness compared with surfacing. It works better both at small and at large unit loads (up to 9 MPa). In the latter case the difference is significant. The profiles of surfaces after the tests demonstrate that molten layers are uniformly, without visible brittle failure. It is conditioned by a high disperse structure, an even distribution of hardening phases and plastic γ -phases. Laser melting increases the wear resistance of alloys of a hypereutectic structure containing excessive diboride crystals. Also the quantity of excessive crystals and their sizes decreases. Although, the absolute hardness values in plasma spraying are much higher than for laser sprayed coatings. The coating structure becomes eutectoid [9]. Laser melted layers, when compared to plasma coatings, show no brittle wear.

The results of wear tests demonstrate that the powder materials on the base of steel do not worsen but surpass self-fluxing nickel-based alloys after laser surfacing. It is possible to explain such an effect by the influence of oxidation on friction and by the formation of secondary structures, which are capable to self-organization.

At sintering of diamond-containing composites in form of blends rollers were formed (Figure 9, a) and it was impossible to control their size, shape and accuracy. The study of the composite beds with diamond grains obtained by means of liquid-phase sintering showed that diamonds rearranged in chains or regular groups (Figure 9, b).

Apparently it is caused by the difference in diamond and binder density. This fact can be considered as a basic effect in the development a production process for single-row multilayer diamond tools.

A new technique of laser thermo-deformational sintering was offered, and it differs from the existing ones by the formation of the working layer in special moulds with help of laser heating and further compacting of crystallized composite with plastic deformation.

The information gained during the experiments resulted in the development of diamond-containing tools and technological equipment.

Two basic sintering schemes are offered in order to manufacture the grinding wheels (Figure 10, a, b).

Axial scheme (Figure 10, *a*) sinters diamonds by means of introduction of powder binder and diamonds into the mould with a wheel case secured in it. Laser beam scans in radial direction with the amplitude that



Figure 9. Laser sintering of powder systems PC12NBKC1 (10 %)--Co (90 %)--diamond (a) with the formation of regular diamond groups (b)





Figure 10. Scheme of axial (a) and radial (b) layer-by-layer thermo-deformational laser sintering of diamond-containing composite on grinding wheels

equals the width diamond-containing layer, than it melts the edges of wheel case and the metallic binding. At the same time the scanning frequency ensures the quasicontinuity of heat flow with intensity required to melt metals. The solidified but still plastic diamond-bearing binder compresses as the mould moves following by the roller that creates necessary geometry of a grinder. Other options of laser sintering could be realized with this scheme as well.

Scheme of radial layer-by-layer building (Figure 10, b) implements the sintering process of onerow multi-layer diamond composites. The powder binder is introduced into the zone of laser heating by means of moving wheel case positioned in-between the immobile matrix. Molten and solidified binder is consolidated by the deforming element. The forming matrix consists of two parts with complicated shape that allows control the height of the sintering layer at the given rotation angle.

Each of the above-mentioned schemes is intended to solve specific technological tasks. They do not exclude but complement each other thus broadening the technological application of these methods.

These schemes were tested at laser sintering of diamond composites on real tools. The results showed that these technologies are very efficient.

CONCLUSION

Short-time (influence time 0.2--0.4 s) laser heating of diamonds AC125 400/315 and AC160 400/315 in air up to the temperatures of 1200--1500 °C does not lead to their thermal destruction and tangible graphitization. The only exception is initially defected diamonds.

Mathematical modeling has shown the temperature distribution in the sintering layer, grinding wheel and adjacent half-matrix thus allowing us to control the process of their forced cooling and, consequently, the quality of laser sintering.

It was proved that it is possible to use wear-proof and durable nickel and cobalt--iron-based powder alloys as a binding material and to use them for the manufacturing of multi-layer diamond cutting tool. The fact of firm covering of diamond grains with binder was established, as well as the formation of the metallic films on the grain surface and consequent formation of ordered grain chains under the influence of laser sintering.

In order to remove the residual thermal stresses that occur at laser sintering and to obtain the necessary configuration and quality of diamond tool it is necessary to combine the laser irradiation process with local deformation of diamond-containing layer in a heated state.

There were also developed means and technological schemes of thermo-deformational sintering of diamond-containing composites in grinding wheels with help of scanning laser beam and layer-by-layer building.

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PECULIARITIES OF BASE METAL PENETRATION IN ARC SURFACING IN LONGITUDINAL MAGNETIC FIELD

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It is shown that in surfacing in constant longitudinal magnetic field (LMF) of 50 Hz frequency depth and area of the base metal penetration reduce if longitudinal component of induction exceeds 50–60 mT level. In case of wire surfacing, rarefaction forms in LMF on the arc axis. Constant and variable magnetic field of 50 Hz frequency exerts braking action on velocity of molten metal flows in the pool, and this is one of main factors that cause reduction of depth of the base metal penetration in submerged arc surfacing with wire.

Keywords: arc surfacing, low-alloy steel, longitudinal magnetic field, induction, weld pool, arc pressure, penetration

In arc welding in longitudinal magnetic field (LMF) such unfavorable phenomenon as reduction of the base metal penetration depth is observed [1], which in regard to the process of submerged arc surfacing with wire is favorable. It is noted in [2] that in submerged arc surfacing with wire increase of induction of LMF of 50 Hz frequency reduces depth of the base metal penetration. Being important for practice of fulfillment of surfacing operations, effect of the base metal penetration efficiency reduction. This work is devoted to specification of effect of the LMF influence on depth and area of the base metal penetration in surfacing and reasons of this influence.

Surfacing was performed on direct current of reverse polarity using the Sv-08GA wire (ferromagnetic) of 4 mm diameter on plates from the 09G2S steel (ferromagnetic) of 20 mm thickness using the AN-348A flux in LMF under the following conditions: $I_{\rm s} = 500$ --550 A, $U_{\rm a} = 30$ --32 V, $v_{\rm s} = 27$ m/h. In all experiments longitudinal component of induction B_z was measured near surface of this plate at distance

from flat end of the electrode to the plate 5 mm. Induction of constant LMF was measured by milliteslameter of the EM-4305 type with Hall sensor with the base 1×1 mm, and variable LMF ---- by milliteslameter of the F-4356 type with Hall sensor with the base 4×4 mm.

For creation of the magnetic field solenoid with ferromagnetic core, installed coaxially with the electrode, was used. Surfacing was carried out under action of constant and variable LMF of 50 Hz frequency. In order to reduce radial components of induction B_r near the electrode in molten metal of the pool, submerged surfacing was also performed using the Sv-12Kh18N10T wire (non-magnetic one) of 4 mm diameter on plates of 20 mm thickness from the same steel using the AN-20S flux under mentioned above conditions. In Figure 1, a, b data of cross sections of beads are presented at surfacing with invariable rate of the wire feeding. At increase of induction B_z of both alternating of 50 Hz frequency and constant LMF, depth of penetration $H_{\rm p}$ reduces to the same degree (see Figure 1, *a*, curves 1 and 2). In addition, in case of using non-magnetic materials, when in zone of the molten metal near the electrode induction com-



Figure 1. Influence of LMF induction B_z on parameters H_p (dark signs) and F_p (light signs): $a - v_{feed} = const$, t, 4 - Sv-08GA of 4 mm, LMF of 50 Hz frequency; 2, 5 - Sv-08GA of 4 mm, constant LMF; 3, 6 - Sv-06Kh19N9T of 4 mm, LMF of 50 Hz frequency; $b - I_s = const$, t, 3 - Sv-08GA of 4 mm, LMF of 50 Hz frequency; 2, 4 - Sv-08GA of 4 mm, constant LMF

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Figure 2. Scheme to calculation of arc gas-dynamic pressure distribution along radius R: a, b — oscillograms of forces F_i of moving at speed of surfacing arc on probe under action and without action of LMF respectively; c — scheme of processing of measurement results (see designations in the text)

ponent B_r is practically absent, depth H_p and area of penetration $F_{\rm p}$ of the base metal also significantly reduce at increase of induction B_z of LMF of 50 Hz frequency (see Figure 1, a, curves 3 and 6). It is characteristic that at induction $B_z \leq 50$ --60 mT depth of penetration practically does not change, and at $B_z \ge$ \geq 50--60 mT it starts to significantly reduce. As far as at switching of LMF reduction of I_s and increase of $U_{\rm a}$ are observed, depositions were also performed, when under action of LMF rate of wire feeding was increased and the same value of current I_s (in all experiments) was preserved. Data on depth H_p and area $F_{\rm p}$ of penetration depending upon the LMF induction (see Figure 1, b) in this case practically coincided with the data, presented in Figure 1, a; established dependences also coincided. For finding out reasons of the observed effects the investigations was carried out, results of which are presented below.

Arc pressure was determined in open arc surfacing with the Sv-08GA wire of 4 mm diameter by means of passage of the arc above tungsten probe of 1.0 mm diameter, axis of which was established at distance 1.5 mm from surface of the plate, and measurement of the force exerted by flows in arc plasma on the probe. Installation for the measurements and methodology of investigation of gas-dynamic arc pressure are described in [3]. It turned out to be possible to fulfill mentioned measurements, because at melting of the electrode extremely fine drops formed, which did not cause shorting of the tungsten probe with the item. Distribution of the arc pressure over radius was obtained after processing of the experimental data on force of the arc action on tungsten probe according to the following methodology.

In the same way as in [3], records of forces in time t, which exerted moving at speed of surfacing arc on the immovable probe, were used as initial information. Typical oscillograms of the arc force action on the probe are presented in Figure 2, a and b, for the process of arcing without action and with action of LMF respectively. Obtained curves were symmetrical. Taking into account scale of image over horizontal (in time) and rate of surfacing, distance between points A and B on these Figures corresponds to the arc diameter D. Due to sufficient symmetry of the curves value of radius R (R = D/2) was determined, and their processing was performed on one of the branches of this curve along radius R.

Arc section (in form of a circle), in which force of the arc action on the probe was measured, was divided into a series of rings, to which corresponded j numbers (j = 1, ..., n; n is the number of rings, Figure 2, c). Force of arc action on the probe F_i , when it was in the *i*-th position, was determined from the oscillograms (Figure 2, a, b), whereby it was assumed that within the whole area of a singled out *j*-th ring gas-dynamic arc pressure p_j was constant. Force, exerted on the probe, is proportional to area of the probe, introduced into the *j*-th ring. In general form equation for determination of the arc force of action on the probe in the *i*-th position has the following view:

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$$F_{i} = C_{d}d_{p}\sum_{j=1}^{i} p_{j}2l_{ij},$$
 (1)

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Figure 3. Distribution of gas-dynamic arc pressure along radius *R* under action of constant LMF (2, 3) and LMF of 50 Hz frequency (4, 5): $1 - B_z = 0$; 2, 4 - 63-65; 3, 5 - 100 mT

where *i* is the probe position number (i = 1, ..., n); d_p is the diameter of the probe, introduced into the arc; l_{ij} is the half of lengths of the probe sections, located in the *j*-th ring at *i*-th position of the probe; $l_{ii} = \sqrt{R_i^2 - h_i^2}$; $l_{ij} = \sqrt{R_j^2 - h_i^2} - \sqrt{R_{j+1}^2 - h_i^2}$ (j < i); $h_i = R_i + R_{i+1}/2$; $R_j = R - R/n(j-1)$; h_i is the distance from the arc section axis to the probe axis in its *i*-th position; R_j is the external radius of the *j*-th ring; C_d is the wire drag factor, which is function of Reynolds number [4]:

Re =
$$\frac{v_{\rm f} d_{\rm p} \rho}{\eta}$$
,

where v_f is the plasma flow velocity; η is the dynamic viscosity of plasma; ρ is the arc plasma density.

Carried out estimations showed that for the arc plasma [5] $10^2 < \text{Re} < 10^4$, whereby value C_d is close to 1 [6]. In the calculations C_d was assumed equal 1.0.

In Figure 2, *c* as an example respective designation are presented for the case of the circle division into five rings. Using program Microsoft Excel pressure p_j in each ring was found. For example, value of pressure p_1 in first (external) ring was determined from equa-



Figure 4. Influence of LMF induction B_z on integral force (1, 2) and mean gas-dynamic pressure of arc (3, 4): 1, 3 — constant LMF; 2, 4 — LMF of 50 Hz frequency

tion, proceeding from (1), having assumed value of force F_1 from Figure 2, *a* at position of the probe at distance h_1 from the arc axis. Then pressure p_2 in second ring was determined, assuming that value of pressure p_1 in first ring is known, etc. Block of calculation formulas for n = 5 has the following form:

$$p_{1} = \frac{F_{1}}{2C_{d}d_{p}l_{11}},$$

$$p_{2} = \left[\frac{F_{2}}{C_{d}d_{p}} - p_{1}2l_{21}\right]\frac{1}{2l_{22}},$$

$$p_{3} = \left[\frac{F_{3}}{C_{d}d_{p}} - p_{1}2l_{31} - p_{2}2l_{32}\right]\frac{1}{2l_{33}},$$

$$p_{4} = \left[\frac{F_{4}}{C_{d}d_{p}} - p_{1}2l_{41} - p_{2}2l_{42} - p_{3}2l_{43}\right]\frac{1}{2l_{44}},$$

$$p_{5} = \left[\frac{F_{5}}{C_{d}d_{p}} - p_{1}2l_{51} - p_{2}2l_{52} - p_{3}2l_{53} - p_{4}2l_{54}\right]\frac{1}{2l_{55}}.$$
(2)

It should be noted that described method allows determining pressure of the electrode metal vapor jets and plasma on the probe, but does not allow obtaining data on excessive pressure of magnetic field inside the arc. Data on distribution of gas-dynamic pressure along the arc radius R (Figure 3) showed that at increase of the LMF induction B_z pressure in axial zone of the arc with a consumable electrode reduces, while in direction of radius it increases, passing through the maximum, which corresponds to data of [7], established for the process of TIG welding in argon atmosphere. Integral force of the arc action and mean gas-dynamic pressure on base metal were respectively determined according to formulas

p

$$F_{\rm int} = \sum_{j=1}^{n} p_j \pi (R_j^2 - R_{j+1}^2), \qquad (3)$$

$$p_{\rm mean} = \frac{F_{\rm int}}{\pi R^2}.$$
 (4)

Integral force of the arc action on base metal (Figure 4, curves 1 and 2) and mean gas-dynamic pressure of the arc (Figure 4, curves 3 and 4) at low values of the LMF induction increase at increase of induction of both constant and variable LMF of 50 Hz frequency. It is characteristic that these parameters stabilize, if the LMF induction exceeds roughly values 50--60 mT.

So, observed reduction of H_p at increase of the LMF induction B_z and especially at B_z above 50--60 mT (see Figure 1) can not be explained by change of the integral arc force or mean arc pressure, because that do not reduce within mentioned range of the LMF inductions. It was assumed that explanation of the fact of H_p reduction at increase of the LMF induction B_z one has to search analyzing phenomena, connected with hydrodynamics of molten metal in the weld pool and, first of all, with braking of this metal flows under action of LMF.

Action of constant LMF on moving from head to rear part of the pool molten metal at velocity v initiates in this metal cross electric field of E intensity (Figure 5) [8]. This process causes occurrence of cross



voltage $U = Eb_p$ (b_p is the pool width). Let us single out tube of current, shown in Figure 5. Voltage U in this tube causes occurrence of cross induced current I. As far as direction of velocity v is perpendicular to direction of vector B_z , one may write for module of the electric field intensity:

$$E = vB_z, \tag{5}$$

$$U = Eb_{\rm p} = IR,\tag{6}$$

where R is the resistance of the circuit, in which current I is shortened (in the base metal).

It follows from formulas (5) and (6) that value of the current is proportional to B_z and v:

$$I = \frac{U}{R} = \frac{Eb_{\rm p}}{R} = \frac{b_{\rm p}B_z v}{R}.$$
 (7)

On singled out tube with current I acts force F [8]:

$$F = IB_z b_p. \tag{8}$$

Using expression (7) and having excluded current, we shall obtain

$$F = \frac{pB_z^2 b_p^2}{R}.$$
 (9)

It should be noted that force F is also directed counter vector v, i.e. reduces velocity of the flow. Considered phenomena in the pool are close in essence to the phenomena, proceeding in MHD-generators, when on a moving in the magnetic field flow of electrical-conducting material (plasma) also acts braking force. Under action of alternating LMF (including the one of 50 Hz frequency) direction of force F at change of direction of the B_z induction does not change. Braking effect preserves in this case as well, but braking force F will pulsate with double LMF frequency. So, under action of both constant and alternating LMF braking force F causes reduction of velocity v of the near-bottom metal flow, increase of liquid interlayer thickness of the molten metal under the arc, and, as a result, reduction of depth and area of the base metal penetration.

Reduction of depth and area of the base metal penetration during surfacing in presence of LMF of 50 Hz frequency, when the item and the electrode are not ferromagnetic, may be also explained by mentioned factor of braking of the molten metal flows.

In case of induction B_z increase of LMF of 50 Hz frequency, outlines of the penetration zone correspond to increasing arc pressure in direction of the radius (see Figure 3), which is manifested in reduction of H_p in center (over axis) of the bead and increase of H_p in peripheral (over bead width) part. Probably, as it was observed in tungsten electrode surfacing in atmosphere of argon [3], density of the arc heat flow along radius under action of LMF also has maximum not in center of active arc spot on the item, but changes in the same way as gas-dynamic arc pressure. It is also probable that in similar way diameter of the arc active



Figure 5. Scheme to calculation of force F, which brakes velocity of flow of metal in pool: 1 — solenoid that generates LMF; 2 — electrode; 3 — molten metal of weld pool

spot on the item also somewhat increases at the LMF induction B_z increase. Comparison of outlines of the metal penetration zone in case of surfacing without LMF (Figure 6, a) and with LMF of 5 Hz frequency and $B_z = 90$ mT (Figure 6, *b*) shows that outlines of this zone do not change, just depth of penetration reduces and width of the bead increases. Under action of LMF with induction $B_z = 150$ mT, H_p reduces in center of the bead and increases in peripheral part (over width) of the bead (Figure 6, c). This is especially manifested in case of using for surfacing of constant LMF with induction $B_z > 110$ mT (Figure 6, d). In similar way changes shape of penetration zone also in case of surfacing with application of constant LMF and of 50 Hz frequency using non-magnetic electrode and item (Figure 6, e). One may assume that a certain contribution into formation of such outline of the penetration zone are made not just by changing radius of current density in anode spot of the arc on the item and gas-dynamic pressure of the arc column plasma, but also by electrode drops. Under action of both constant and variable LMF of 50 Hz frequency drops on end of the electrode acquire pulse of force and velocity, directed along tangent line to the cir-



Figure 6. Sections of deposites, carried out at $v_{\text{feed}} = \text{const: } a - \text{surfacing without LMF; } b - \text{LMF of 50 Hz frequency, } B_z = 90 \text{ mT; } c - \text{the same, } B_z = 150 \text{ mT; } d - \text{constant LMF, } B_z = 110 \text{ mT; } a - d - \text{Sv-08GA wire and plate from 09G2S steel (} I_s = 500 - 550 \text{ A}, U_a = 30 - 32 \text{ V}); e - \text{LMF of 50 Hz frequency, } B_z = 70 \text{ mT, wire and plate are from non-magnetic materials}$

cumference and downwards, i.e. they move over cone surface. The drops enable by their heat content and pulse growth of the metal penetration on bead edges. However, main contribution into general reduction of depth H_p and area F_p of the base metal penetration under action of LMF is made by braking action on velocity of the near-bottom flow of metal in the pool from head to rear part of the latter axial (longitudinal) component of the LMF induction. Braking action on velocity of metal flow in the pool is proportional to square of the LMF induction B_z and velocity of these flows.

CONCLUSIONS

1. In submerged arc surfacing with wire, depth of the base metal penetration reduces if value of longitudinal component of constant and variable LMF of 50 Hz frequency exceeds level of about 50--60 mT. This is observed in case of using wires and items both from ferromagnetic and non-magnetic materials.

2. Reduction of the base metal penetration efficiency in surfacing under action of constant and variable LMF of 50 Hz frequency is determined not just by change of pressure distribution over radius of the arc and the arc radius, but mainly by braking by this field of the near-bottom flow of molten metal from head to rear part of the pool.

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ON THE INFLUENCE OF SMALL PARAMERTERS ON MIG/MAG WELDING STABILITY

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The paper gives the results of analytical and experimental investigation of dynamic processes running in the welding circuit, and more precisely defines the area of asymptotic stability of steady-state arc welding modes.

Keywords: automatic arc welding, arc discharge inertia, stability, transient processes

Provision of welding process stability against external impacts and parametric disturbances is known to be one of the basic requirements in development of advanced arc welding technologies and appropriate welding equipment. Stability of the steady-state modes of arc welding is usually verified using the known criterion [1--7], according to which these modes will be stable, if a certain relationship between the slopes of volt-ampere characteristic of the welding arc and current source powering the arc is fulfilled in the working point.

This criterion was established at some time in study [3] at consideration of a simplified mathematical model of the processes running in the welding circuit. This model, however, does not take into account the so-called small parameters, in particular, inductance in the welding current circuit and time constant of the arc characterizing its inertia properties. It is clear that allowing for small parameters in a mathematical model leads to increase of the order of its differential equations, and this gives rise to the question of how well substantiated is the use of the above criterion for

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verification of welding process stability. Hence, the need to study the refined mathematical model and clarify the issue for discussion: should the influence of small parameters on arc welding process stability be taken into account or is it negligible? This work describes the results of investigations aimed exactly at clarification of the posed questions.

Mathematical model. Let us consider a system of equations describing according to [1, 4, 5, 8--13] the processes in the welding circuit at automatic MIG/MAG welding:

$$\frac{d\lambda}{dt} = M(\varepsilon + i_0) - \upsilon_e, \tag{1}$$

$$L\frac{d\varepsilon}{dt} + (R - S_s) (\varepsilon + i_0) = u_* - u_a, \qquad (2)$$

$$u_a = u_0 + E(\lambda + l_0) + S_a(\varepsilon + i_0),$$
(3)

where $\lambda = l - l_0$, $\varepsilon = i - i_0$ are the deviations of current values of arc length l = l(t) and welding current i = i(t) from l_0 and i_0 values in the steady-state mode; $M \equiv \partial v_m / \partial i$ is the steepness of electrode melting characteristic in point $i = i_0$; v_m is the electrode melting rate; v_e is the electrode feed rate relative to torch nozzle; L is the welding current inductance; R is the





total resistance of conducting wires, electrode extension and sliding contact in the torch nozzle; $S_s = = \partial u_s / \partial i$, $S_a = \partial u_a / \partial i$ are the tangents of VAC angles of inclination of welding current source and arc in working point $i = i_0$; $E \equiv \partial u_a / \partial l$ is the electric field intensity in the arc column; u_s is the voltage at output terminals of welding current source; u_* is the opencircuit voltage; u_a is the arc voltage; u_0 is the sum of near-electrode voltage drops; t is the current time.

System of equations (1)--(3) is a somewhat refined mathematical model, which allows for inductance of welding current source, but still does no allow for dynamic properties of the arc proper. Further refinement of this mathematical model can consist in replacement of algebraic equation (3) by differential equation

$$\tau \frac{du_a}{dt} + u_a = u_0 + E(\lambda + l_0) + S_a(\varepsilon + i_0), \qquad (4)$$

where $\tau > 0$ is the parameter characterizing arc discharge inertia.

It should be noted that the used in (3) and (4) approximation of real dependence $u_a = u_a(i, l)$, which is indeed of a distributed nature [14], is quite admissible in this problem, as here we are interested in the so-called microprocesses in the arc proper, and it is considered from the macroscopic viewpoint as an element of welding current circuit differing by some generalized inertia properties.

Solving the system of equations (1), (2) and (4) for $\lambda(t)$ variable at $v_e = \text{const}$, as a result we obtain one differential equation

$$\tau T_e T \frac{d^3 \lambda}{dt^3} + T(\tau + T_e) \frac{d^2 \lambda}{dt^2} + T_s \frac{d\lambda}{dt} + \lambda = 0, \quad (5)$$

where T_e , T, T_s are the time constants determined by the following relationships:

$$T_e = \frac{L}{R_*}, \quad T = \frac{R_*}{EM}, \quad T_s = \frac{R_{**}}{EM},$$
 (6)

$$R_* = R - S_s, \quad R_{**} = S_a - S_s + R. \tag{7}$$

Equation (5), similar to equations (1)--(3), was derived in the assumption that welding is performed by a long arc, and the distance between the edge of the current conducting nozzle and free surface of the weld pool does not change, and condition 0 < l(t) < $< l_k$ is certainly fulfilled, where l_k is the critical arc length l(t) at which its breaking up occurs. At $l(t) \ge$ $\ge l_k$ and l(t) = 0 the above equations are meaningless, as they describe the processes in the welding circuit only at arcing.

Hence, a rather strict limitation on absolute value of deviation $\lambda(t)$ connected solely to arc welding specifics:

$$|\lambda(t)| < \delta, \quad \delta = \min \left\{ l_0, \left(l_k - l_0 \right) \right\}. \tag{8}$$

Right-hand relationship (8) geometrically assigns a band of 2δ width, beyond which $\lambda(t)$ variable should not go.

Stability analysis. From the theory of stability it is known [15--20] that $\lambda(t)$ transient process in the case of fulfillment of condition (8) will certainly attenuate ($\lim_{t \to \infty} \lambda(t) = 0$) if coefficients of equation (5)

are positive, i.e.

$$\tau > 0, \ T > 0, \ T_e > 0, \ T_s > 0, \ (9)$$

and the following equation is fulfilled:

$$T_s(\tau + T_e) > \tau T_e. \tag{10}$$

As L > 0, E > 0 and M > 0, then, according to (6) and (7), the three latter inequalities (9) will be in place, if $R_{**} > 0$, i.e. if under (7) the following condition is fulfilled:

$$S_a - S_s + R > 0. (11)$$

We will present relationship (11) in the dimensionless form:

$$\mu + 1 > 0, \tag{12}$$

where μ is the parameter found from the formula

$$\mu = \frac{S_a - S_s}{R}.$$

Inequalities (11) or (12) provided condition (8) is fulfilled, essentially are a criterion of the so-called conditional asymptotic stability of arc welding process [19], described by a simplified model (1)–(3).

Let us now consider inequality (10), which is due to allowing for arc discharge inertia in the mathematical model. Direct verification of this inequality is made difficult by the fact that valid data on the real numerical values of time constant τ included into expression (10) are, as a rule, absent. Let us avoid that in the following manner. Let us express inequality (10) as

$$T_s > \frac{\tau T_e}{\tau + T_e} \tag{13}$$

and consider its right-hand part. As $\tau > 0$ and $T_e > 0$, it is obvious that

$$\frac{\tau T_e}{\tau + T_e} < T_e$$

irrespective of τ . Therefore, condition (13) and, hence also (10), will be certainly fulfilled, if

$$T_s > T_e. \tag{14}$$

Thus, inequality (14) is a sufficient condition of asymptotic stability of the dynamic processes running in the welding circuit.

Let us write relationship (14) allowing for (6), (7) in expanded form





$$R_*R_{**} > LEM. \tag{15}$$

For convenience of comparison, let us replace R_*R_{**} in the left-hand part of inequality (15) by $R_{**}^2 \approx R_*R_{**}$. Using dimensionless values

$$\mu = \frac{S_a - S_s}{R}, \quad \theta = \frac{\sqrt{LME}}{R} \tag{16}$$

we obtain a criterion of stability (14) in the following form:

$$\mu + 1 > \theta. \tag{17}$$

Comparing formulas (12) and (17) it is easy to see that they differ by their right-hand parts. It is obvious that criterion (17) has an essential advantage, as it allows a more precise determination of the region of asymptotic stability than criterion (12) does. The stability region, according to (17), (16), depends not only on the slope of VAC of the arc and welding current source in the working point, but also on inductance L of welding current source, steepness of electrode melting characteristic M, electric field intensity in the arc column E and total resistance R of current-conducting wires, electrode extension and sliding contact in the torch nozzle. It should be noted that criterion (17) has the same form as the criterion earlier obtained by us based on Silvester inequalities for the non-stationary case, when studying the influence of electric filed fluctuations in the arc column on the stability of arc welding process [21].

Figure 1 shows the region of asymptotic stability based on criteria (12) and (17) in the space of two parameters μ and θ . The limits of stability in Figure 1, *a* (simplified model) are $\theta = 0$ and $\mu = -1$ lines. In Figure 1, *b* (a model allowing for small parameters) stability limits fall on lines $\theta = 0$ and $\mu = \theta - 1$. As we can see, in the refined mathematical model (Figure 1, *b*) the stability region becomes narrower with increase of θ parameter.

From the second relationship (16) it follows that θ values rise at increase of inductance value *L*. It means that at an essential increase of *L* the stability region (Figure 1, *b*) can become markedly narrower. This fact was noted already in [1], where it was reported, in particular, that in addition to the known stability condition $\mu > 0$, also a condition limiting

inductance L, should be defined. This condition now follows directly from expressions (17) and (16):

$$L < \frac{\left(S_a - S_s + R\right)^2}{ME}.$$
(18)

It may turn out, however, that the values of welding current source inductance L are selected proceeding from, for instance, the technological or other considerations. In such a case, only one controllable parameter remains for ensuring a reliable stability of the steady-state mode, namely tangent of the angle of inclination of static VAC of welding current source S_s , at selection of which it is desirable to take into account limitation (18).

Thus, fulfillment of condition (17) and (8) ensures attenuation of the transient processes running in the welding circuit. However, the following question may be raised: will parameters t and L influence the transients quality and to what extent? To clarify this issue computer simulation was performed of the dynamics described by equations (1), (2) and (4). The following values of contour parameters and welding mode were used: $L = 7 \cdot 10^{-4}$ H, M = 0.38 mm/(s·A), E == 2 V/mm, R = 0.015 Ohm, $S_s = -0.02$ V/A, $S_a =$ = 0.005 V/A, $i_0 = 132$ A, $v_e = 50$ mm/s, H = 17 mm, where H is the distance between the end face of the current conducting nozzle and free surface of the weld pool.

Jump-like variation of voltage u_s on the output terminals of welding current source was considered as a disturbance:

$$u_{s}(t) = \begin{cases} 21 \text{ V at } 1.5 \le t < 2.3 \text{ s}, \\ 25 \text{ V at } 2.3 \le t \le 2.8 \text{ s}, \\ 21 \text{ V at } t > 2.3 \text{ s}. \end{cases}$$
(19)

Results of computer simulation are given in Figure 2. Figure 2, *a* shows actually two $\lambda(t)$ graphs obtained at different values of parameter τ : $\tau_1 = 1 \cdot 10^{-2}$ s and $\tau_2 = 1 \cdot 10^{-5}$ s. But these two graphs practically coincide, although time constants τ_1 and τ_2 differ from each other by two orders of magnitude. This may lead to the conclusion that parameter τ does not have any tangible influence on the transient processes in the welding circuit.

Figure 2, *b* shows graphs $\lambda(t)$ obtained at $\tau = 1.10^{-3}$ s and two different inductance values.



Figure 1. Region of asymptotic stability without allowing (a) and allowing (b) for small parameters

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Figure 2. Results of computer simulation of transient processes in the welding circuit: a, b — see the text

Curves 1 and 2 characterize the reaction of arc length on disturbance (19) at $L_1 = 7 \cdot 10^{-5}$ H and $L_2 = 7 \cdot 10^{-3}$ H, respectively. Let us compare curve 1 in Figure 2, b with the curve in Figure 2, a. These curves do not differ from each other, although $L_1 = 7 \cdot 10^{-5}$ H is by an order of magnitude lower than $L = 7 \cdot 10^{-4}$ H. Therefore, small inductance values do not have any noticeable influence on $\lambda(t)$ transient process. Comparison of curves 1 and 2 in Figure 2, b shows that a considerable increase of inductance ($L_2 = 7 \cdot 10^{-2}$ ---- is by an order of magnitude larger than L) leads to an essential change of the nature of curve $\lambda(t)$.

Here, we should mention one circumstance. As is seen from Figure 2, b, inductance increase leads to an increase not only of the transient process time, but also of absolute value of deviation of arc length $|\lambda(t)|$ from its steady-state value l_0 . It is obvious that condition (8) ---- $|\lambda(t)| < \delta$ ---- can be disturbed at a certain value of L (Figure 2, b). This may lead to development of unstable modes with quasiperiodic extinctions of the arc or undesirable short-circuiting in automatic arc welding (depending on the position of band δ = = min $\{l_0, (l_k - l_0)\}$ relative to l_k).

CONCLUSIONS

1. Results of analysis and numerical simulation of the refined mathematical model are indicative of the fact that the arc discharge inertia practically does not have any noticeable influence neither on stability of the processes running in the welding current circuit, nor on transient process quality, and in most cases it may be neglected.

2. In order to assess the stability region, it is rational to use criterion (17), which establishes its dependence not only on the angle of inclination of the VAC of the arc and welding current source in the working point, but also on other parametes of the welding circuit.

3. Unstable modes of automatic MIG/MAG welding because of quasiperiodic breaking up of the arc or undesirable short-circuiting of the arc gap, may arise also at fulfillment of the conditions of stability, if this leads to disturbance of condition (8), caused, in particular, by too high inductance of welding current source.

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EXPERIMENTAL FACILITY FOR RESEARCH ON PULSED LASER-MICROPLASMA WELDING^{*}

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Functional structure of a technological complex for hybrid (laser-microplasma) welding with application of a solid-state Nd:YAG laser, operating in a repetitively pulsed mode with wide range of energy and pulse shape, and integrated direct-action coaxial plasmatron is designed. Suggested algorithms of synchronizing operation of pulsed heat sources for study of joint action from positions of required welding properties achievement give the opportunity to obtain experimental results of laser-microplasma welding process with mild steel for automobile industry, aluminium alloy processing, and nanomaterials sintering.

Keywords: hybrid welding, microplasma arc, laser radiation, pulse, experimental facility, arc current, shape of penetration

Potentialities of the modern industry significantly depend on progress in technologies of material processing and joining. It is hard to overestimate the importance of creation and development of welding, cutting or surface modification techniques, based on application of arc discharge plasma as the most cheapest and available energy source. At present, the development of plasma technologies requires [1] concentration of energy in the plasma arc and stabilization of discharge at simultaneous increase of welding process productivity.

The alternative way, widely developed and used in the industry, is based on application of laser radiation as the only thermal source, so jointly with other traditional heating techniques [2, 3], including arc discharge, HF wave, or light irradiation. Due to high concentration of energy in the beam and opportunity of local action, a laser provides high productivity, deep penetration and high accuracy in adjustment of shaping parameters and weld metal properties.

Despite obvious advantages, opportunities of the laser as a thermal source are significantly limited for some applications. First of all, it is due to low efficiency of metal heating by laser radiation because of its high reflectivity at radiation wavelength, typical for the most of technological lasers. Another factor, decreasing efficiency of power laser application, is induced by laser radiation plasma over the metal surface, which reduces the energy part, contributed by a laser beam into a processed object.

A new approach to solution of the above-mentioned problems is based on use of combined laser-arc and laser-plasma processes, essence of which is in joint action on processed object of laser radiation and arc plasma [4, 5]. At practical realization of the combined processes both thermal sources act on metal surface inside the common heating zone. Heating of metal by the electric arc leads to a rise of its temperature and, as a result, to increase of absorption of laser radiation. In turn, a small spot of laser beam creates on the metal surface a definitely localized zone with increased concentration of free electrons that raises the efficiency of the discharge. As a result, application of two different sources can lead to occurrence of positive synergetic effect, which reveals itself in action efficiency rise both for laser radiation and arc discharge [6].

To study the process of hybrid repetitively pulsed laser-arc processing of materials, an experimental facility, containing a pulsed microplasma source and a laser with a controllable pulse shape and beam coaxial to the arc, was created. It has been designed to consider the following tasks:

• estimate of potential synergetic effects in studies on penetration with variation of delays between the pulses of different sources;

• preliminary investigations on action of spatial and temporal parameters of the laser radiation and arc inter-positioning on processing efficiency;

• revision of technical requirements to hybrid heating source;

• design of functional schematics and structure of equipment for repetitively pulsed laser-arc material processing;

• applied technological studies of hybrid welding for refinement of the functional structure and parameters of the complex;

• formation of the technical requirements to subsystems of the complex;

• study of technological potentialities of the equipment for hybrid material processing;

• experimental application of the technological process and equipment.

Actuality of the development is defined by the following key issues:

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Figure 1. Construction of the integrated plasmatron with focusing system 1, alignment mounting unit 2, channel for input of plasma gas 3, collet 4, cathode unit body 5, isolator 6, plasmatron body 7, gas splitter 8, input nozzle for protection gas 9, thermocathode 10, porous filling 11 and plasma-shaping nozzle 12

• creation of a low-power technological facility for laser-arc processing of materials ensures a wide variation range of processing capabilities for formation of the required properties of the weld metal;

• joint action on metals can lead to increase in efficiency of energy transfer for both laser, and arc heating sources;

• maximum depth of penetration into metals, defining the thickness of welded parts, is more than one and a half higher in comparison with the only arc source that results in practical productivity doubling [7].

Technical parameters and structure of the laserarc facility. As the laser base of the technological equipment for hybrid material processing, we have chosen a laser technological complex, intended for cutting of thick metal sheets and deep engraving. Main



Figure 2. Functional schematics of the laser-arc facility (for designations see the text)

parameters of the prototype laser technological equipment are as follows:

Laser type	Nd:YAG
Operation mode	repetitively pulsed
Pulse shape	digitally controlled,
variable	in a wide duration range
Operating frequency (limited by 12	0 W
average power level), Hz	up to 100
Average power, W	up to 120
Pulse energy (limited by 120 W ave	erage
power level), J	up to 3

As a source of electric arc, we have chosen the direct-acting plasmatron developed in the E.O. Paton Electric Welding Institute [8]. It has changeable plasma-shaping copper nozzle with output hole diameter of 1 or 1.5 mm, intended for limitation of lateral dimensions and spatial stabilization of the arc discharge. Construction of the integrated plasmatron allows us to bring focused radiation of the pulsed laser in the welding zone coaxially with the plasma beam.

Figure 1 shows the schematics and design of the plasmatron. It is designed to provide several modes of arc discharge, including CW and repetitive with straight and reversed polarity of the pulses at total power of up to 600 W. Figure 2 presents the functional schematics of the laser-arc facility. The laser complex consists of laser oscillator 1, beam-delivering system 2, and laser supply unit 4. The plasmatron complex consists of pulsed plasmatron 5, plasmatron supply unit 7, and gas-delivering system 3. The integrated plasmatron 5 is mounted to the laser focusing system 2, and this fixation provides alignment of the beams. The control system 10 and clock system 9 ensure synchronization of the laser pulses with arc discharge pulses and two-axis table 8 motion for realization of hybrid mode processing of workpiece 6. Appearance of the facility is given in Figure 3, and operation of the complex is shown in Figure 4.

Results and discussion. Figure 5 shows penetration of welded samples obtained using stainless steel and the laser-arc facility. The technical parameters of



Figure 3. Appearance of the facility for research on hybrid welding processes



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Figure 4. Different modes of laser-arc facility operation



Figure 5. Weld structure at addition of laser radiation to the arc discharge in penetration welding stainless steel 1 mm thick at average power of synchronized laser pulses of 35--40 W (for the rest parameters see Table 1)

the arc source, corresponding to the presented welds, are given in Table 1. Motion of the table has been leftward, and it is seen from the picture that addition of laser radiation to arc plasma action results in noticeable weld structure.

The preliminary results of experimental studies have shown noticeable refinement of the weld (shape stabilization) in comparison with the arc welding, even when a part of the laser radiation power is only 10-15 % of total input power.

To analyze the mechanism of this phenomenon, a series of experiments have been performed. First of all, electric measurements show reduction of the electric arc discharge voltage at laser plasma emerging (Figure 6). It can be considered as indirect qualitative proof of discharge efficiency increase at simultaneous usage of both tools. This effect theoretically described in [9].



Measurement of the arc discharge gap resistance in presence of laser radiation has been produced with low supply voltage of 5 V and limiting ballast resistor of 1000 Ohm to prevent arc discharge initiation (Figure 7). It is important to note that resistance of the gap during the laser pulse is even lower than that of ballast resistor (with the voltage on the gap below 2 V). It means that electron concentration in the laser torch is even higher than in the arc discharge.

Interaction effects for combination of laser radiation and arc plasma in welding of aluminum are even more noticeable due to surface alumina film, preventing stabilization of the anode spot for the arc discharge of the straight polarity. The laser energy in this case can be used for removal of the film in the limited zone of laser focal spot. Figure 8 presents the results of the experiments on weld penetration for aluminum samples with arc pulsed discharge action and joint arc and laser action. Parameters of laser pulse in the experiments are presented in Table 2.

Mode	Arc current, A	Pulse duration, ms	Pause duration, ms	Frequency, Hz	Welding speed, m/min
Pulsed	18	10	7	60	0.18
Pulsed					0.12
CW	16				0.30

Table 2. Technological laser parameters for welding of aluminum samples with penetration at frequency of 70 Hz and pulse duration of 1 ms according to Figure 8

Sample	Current, A	Average power, W	Pulse power, W	Intensity, W/cm ²
а	100	61	870	$5 \cdot 10^6$
b	150	72	1000	$5.8 \cdot 10^{6}$
с	200	80	1150	$6.5 \cdot 10^{6}$
d	250	87	1250	$7 \cdot 10^{6}$





Figure 7. Schematic (a) and measurement (b) on resistance R_x of the arc discharge gap in laser torch at supply voltage $U_0 = 5$ V and ballast resistor $R_0 = 1000$ Ohm

Variation of the laser power allows us to define experimentally the threshold intensity of laser radiation required to remove the oxide film. At excess of this value, arc discharge jitter disappears, and reliable binding of the anode spot to the laser focal spot takes place, as it is seen in Figure 8, d. The previous image (see Figure 8, c) corresponds to the threshold value of laser intensity and gives, according to Table 2, the required intensity for removal of the alumina film, equal to $6.5 \cdot 10^6$ W/cm².

Analysis of the film destroy mechanism has shown that two variants are possible: first of them is connected with absorption in the film itself with further evaporation, and the second one is due to absorption in the aluminum substrate (Figure 9). To estimate the temperature of the aluminum heating, it is necessary to solve the problem of thermal conductivity. In our case, lateral conductivity can be neglected, so the

Straight polarity arc Hybrid arc-laser discharge

Figure 8. Weld penetration on aluminum surface in conditions of straight polarity arc and hybrid arc-laser discharge

temperature rise for the rectangular laser pulse is defined by the following expression [10]:

$$T = \eta I \, \frac{\sqrt{\chi t/\pi}}{\lambda},\tag{1}$$

where η is the absorption coefficient of the aluminum surface; *I* is the power density of laser radiation; χ is the temperature conductivity; *t* is the time of action (pulse duration); and λ is the thermal conductivity.

In supposition that all energy goes on heating till the evaporation temperature, it is possible to obtain the expression for the required laser intensity. In this case, one obtains estimation from below, since it does not take into account losses of energy on melting, partial evaporation, needed to break the film, and weak lateral conductivity. The value of laser intensity, calculated by formula (1), is $4.5 \cdot 10^6$ W/cm², that well agrees with the experimentally received results.

Estimation of the second possible mechanism due to absorption in the film itself according with the expression, taking into account double pass of radiation through the film because of reflection from the aluminum surface

$$I = \frac{c\rho T}{\alpha t (2 - \eta)},\tag{2}$$

where *c* is the heat capacity; ρ is the density; and α is the absorption coefficient, gives the significantly lower



Figure 9. Processes of radiation absorption, reflection and heat transfer at removal of the oxide film on the aluminum surface $\$



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value of $2.5 \cdot 10^6 \text{ W/cm}^2$. This fact can be explained by effective heat transfer from the film to the aluminum substrate due to smaller thickness of the film that ensures sufficient conductivity in the millisecond range.

Pulsed mode allows us to plan experiments on quantitative study of the potential synergetic effect. To exclude mutual influence of laser radiation and arc plasma, we can use temporal delays between their pulses so that the studied object starts to act on metal surface just after completion of the auxiliary action. In this case, estimation of the delivered power and effect value must be based on a precise technique, and its development is one of the main scientific problems of the project.

CONCLUSIONS

The developed hybrid laser-arc facility has the following peculiarities:

• laser source provides radiation pulses with peak power up to 4 kW and intensity up to $10^7 \ W/cm^2$ in the processing zone;

• arc source supplies four different modes, including pulses of reversed polarity, with repetition rate up to 70 Hz and pulse current up to 30 A;

• clock system ensures variation of a delay between arc current pulse and digitally shape-controlled laser radiation pulse;

• coaxial scheme of laser radiation delivery and arc current channel results in maximum efficiency of torch--arc interaction;

• construction of the plasmatron makes it possible to regulate the contents of the gas mixture and separate delivery of plasma and protection gases;

• two-axis table gives the opportunity of automatic adjustment of workpiece motion parameters in the linear speed range of 0.006-0.6 m/min with inde-

pendent vertical displacement of the plasmatron over the surface being processed.

The initial experiments performed with its application show high potentialities of the equipment for study of hybrid welding mechanisms and development of the technological modes. The obtained results prove the opportunity of the essential influence of laser and arc pulse interaction on the molten zone shape. Classification of the factors and their quantitative estimation require further studies with application of metallographic analysis for investigation on the pulsed hybrid welding processes.

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ON THEORY OF SOLUTION OF INVENTION TASKS

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The theory of solution of invention tasks includes method of inversion (re-arrangement) of the known elements of devices to obtain new properties. Below the method of «turning inside out» is suggested, which expands or complements capabilities of the inversion.

Keywords: welding production, inductor, proportioner, switch, manipulator, inventions, method of inversion

Essence of the method may be illustrated on example of using a skin with a fur as a fur coat. Borrowed from animals idea of the warm cloth saved many people from cold. However, wetted by rain fur loses its heatinsulating properties, and the fur coat does not heat in the same way as the dry one. Besides, problems of drying and maintaining of heat-insulating properties of stuck together fur are added. That's why people decided to turn skin of the fur coat inside out. Rain, draining off dense surface of the skin, leaves the fur dry. The new turned inside out item had evident advantages. In this way tanned sheepskin coat was invented.

This method may be used in engineering, due to which not just new quality may be obtained, but unexpectedly functional possibilities of items and devices may be expanded.

Example 1. Use of the method in a static item. *Multi-seat inductor for heat treatment, brazing and surfacing*. Such inductors are sometimes called multi-





eye, multi-place, multi-position, and even multi-cell ones.

Multi-seat inductors, which allow simultaneous processing of several parts on the same unit, are used for increasing productivity. Design of the known devices is presented in Figure 1. Here in the induction contour holes (seats) are made for placement into them of the parts being heated. The seats are interconnected by slots, located inside the inductor contour along the line of placements of the parts, whereby extreme positions have one slot in contrast to medium ones each of which has two slots. Number of slots in the seats being different, angle of coverage by current of the items, located in extreme positions, is somewhat higher than in medium ones. That's why the latter ones are heated slower. If number of the seats is odd, in addition a hole with three slots appear.

For equalization of heating the following methods are used:

• extreme holes are made of somewhat bigger diameter, thus reducing efficiency factor of the process, because efficacy of the process is maximum at minimal spacing between the item and the inductor;

 \bullet on medium positions additional Π -like magnetic conductors are installed, which complicates the device and not always is acceptable because of design peculiarities of the items;

• in extreme seats additional slots are made, which equalize conditions, but increase time of heating of the parts.

And now let us try turn the inductor inside out. Due to the method of «turning inside out» one may get a multi-seat inductor, providing equal conditions of heating for all parts, increase of efficiency, and simplification of the design.

In Figure 2 the inductor is shown, in which the holes are arranged in two rows in chess-board order, and slots of all holes are directed not inwards, but outwards of the inductor perpendicular to the direction of the rows. Such design ensures the same angle of coverage by current of the items, located in all holes and allows using the inductor with maximum efficiency factor by arranging minimal spacing between the items and the inductor. Here, of course, it was not sufficient just to turn inside of the closed contour with semicircles of the holes outside; it was necessary to find form of arrangement of the holes, which ensures in deployed contour the conditions similar to a simple single-seat inductor. But impetus to



Figure 2. Scheme of multi-seat inductor according to USSR author's certificate No 909852

the search was given by the method of turning inside out.

Example 2. Use of the method in the proportioning device. Disk proportioners for feeding of powder masses are known (Figure 3), which consist of the hopper, a rotating disk (a plate) and a scraper. Functional possibilities of such design are demonstrated on the scheme, and they consist in feeding of the powder into a certain point. Regulation of amount of the powder is mainly performed by change of revolutions of the disk and spacing under neck of the hopper. The powder is thrown by the scraper outside the limits of external diameter of the disk.

Is it possible to turn inside of such proportioner outside and what will be the result?

For implementation of the idea it was necessary not just to solve problem of the doming, i.e. to overcome inclination of the free-flowing materials to form hang-ups above the holes, but to use them for proportioning. In this way method of feeding free-flowing materials was born, which used a «running hang-up» (Figure 4).

In Figure 4, a an initial position of the process is shown, when the powder material 5 formed a hang-up 6 above the hole. In Figure 4, b beginning of proportioning is shown, when due to movement of the powder relative the hole 3 in the hopper bottom 2 the hang-up 6 falls. In Figure 4, c example of uniform movement of the powder relative the hole and its filling through it on the object of proportioning is shown, and in Figure 4, d formation of a new hang-up 6 after stopping of movement of the blades relative the hole in the hopper bottom --- end of the process ---- is shown.

General principal scheme of the proportioner design is presented in Figures 5 and 6. The proportioner has a hopper 1, a replaceable bottom 2 with holes 3, and located above the bottom blades 4, which may





Figure 4. Method of proportioning «running hang-up» according to USSR author's certificate No 134992 (for *a*-*d* see the text)

move parallel to the bottom. The hopper contains powder material 5 ---- the subject of proportioning.

Design of the proportioner repeats the disk design exactly in the opposite way. Powder from the hopper is not emptied through the neck on the disk, but is located on the disk, which is immovable and closes the hopper from below. The scraper is moveable and is made in the form of several radially diverging from the hopper axis blades, connected with the rotation drive.

In Figure 5, *b* the section over bottom holes is shown, on which one may see mutual arrangement of the blades and the bottom holes during proportioning. In Figure 5, *a* an example of the powder flux proportioning through round holes on the annular part for brazing is shown. The holes may be arranged over two concentric circumferences, and object of proportioning may rotate for uniformity of the powder distribution along contour of filling.

In Figure 6 an option of a proportioner with slot holes in a replaceable bottom is shown.

In Figure 7 forms of slots and configuration of a form of the material filling on objects of proportioning are shown. The arrows show direction of movement under the proportioner of the object, on which a free-flowing material is fed during proportioning.

The proportioner operates as follows. After switching of the drive the blades 4 start rotating around a common axis uniformly and parallel to the bottom 2, whereby powder material 5 moves relative holes 3 and the hang-ups 6 fall (see Figure 4, b). Uniform movement of the blades and the powder relative the holes cause due to force of gravity pouring of the powder 5 through the holes (see Figure 4, c), whereby hangups do not have time to be formed, because their support part, hanging above the surface, continuously falls. After termination of movement of the blades 4



Figure 5. Proportioners according to method of designing «turning inside out» (for *a*, *b* see the text)

hang-ups are again formed above holes 3 and feeding of the powder material 5 stops (see Figure 4, d).

Shape of the slots allows feeding different amount of powder on figured surfaces of the object of proportioning (Figure 7). Such design ensures wide functional possibilities of the proportioner and increases productivity of the process due to the fact that more material is fed through the slots than through round holes, and shape of the slots ensures wide range of figured filling.

Bending of the blades relative a shape of the slots ensures that the blade does not completely cover the slot at any moment even for a short time when it is above the profile of the hole.

Example 3. Use of the «turning inside out» method in development of a small-size mechanism. A switch with a cylindrical cam and method of its manufacturing. The invention relates to the mechanisms, in which it is necessary to switch over position of separate parts, for example, to gear boxes of the reducer type with a changeable reduction (transmission) ratio, in which revolutions of the outlet shaft are regulated, in particular, to manipulators of welding production, but it may be also used in other mechanisms, where the need occurs in small-size switches.

In known designs of switches with a cylindrical cam the cam slot is, as a rule, arranged on the side of external surface of a cylinder. Such and similar mechanisms are rather widespread, and elements, which interact with the slot, are always located outside limits of the cylindrical cam. Such design has big overall dimensions and limits possibility of development of small-size devices.

The task of the invention was to expand functional possibilities by reduction of overall dimensions of the switch. Essence of the invention consists in the fact that in the switch with a cylindrical cam, which has a slot arranged on external surface of the cylindrical surface, and an element, which interacts with the slot, the cam, according to the invention, is made with an internal cavity that is concentric in relation to the



Figure 6. Proportioner of Mozzhukhin according to patent of Ukraine 27107 (for 1-5 see the text)



INDUSTRIAL



Figure 7. Forms of slots and configuration of form of material filling on objects of proportioning

cam cylinder, and the slot is arranged on the side of the internal cylindrical surface of the cam cavity, whereby the element, which interacts with the slot, is located inside the cam and does not extend over its dimensions in radial direction, i.e. the known design is turned inside out.

Essence of the invention also consists in the method of manufacturing of the switch with a cylindrical cam. It envisages manufacturing of a cylinder with a slot first on external side, then a bushing with a diameter of the radial hole not less than that of the slot width is pulled on it, and the bushing is welded to the cylinder. Then a concentric hole with greater diameter than internal diameter of the slot is made in the cylinder in such way that the slot opens into cylindrical cavity of the hole. After this a shaft with a threaded seat for the rod is installed into the hole, the seat is set opposite the hole in the bushing, and a pin is screwed into the shaft through it. In this way the element is manufactured, which interacts with the slot, is located inside the cam, and does not exceed its overall dimensions in radial direction.

General principle scheme of the switch design is presented in Figure 8, in which as an example is shown a four-stage gear box 1 with a switch 2 and forks 3. In Figure 8 the section of the box is presented, on which a worm shaft is shown, from which revolutions of the engine are transmitted to the shafts 5 and 6 through gear clusters 7 and 8 and area 1. Switch 2 is connected with shaft 9, which has pins 10, located in slots of cylindrical cams of the fork hubs 3. Forks 3 are located in slots of couplings 11 that have holes 12





Figure 9. Stage-by-stage scheme of manufacturing of switch



Figure 10. Table manipulator OB1020

opposite fingers 13 that are rigidly fixed in gears 8. Couplings 11 are fitted on shaft 5 on key 14 and may move along the said shaft. Gears 7 and 8 are fitted on shafts on bearings and may freely rotate around them. The developed view of the cylinder cams with slots is given, on which shift of active zone of the cylindrical cams relative each other for sequential switching onoff of respective gears in case of the reducer reduction ratio change is shown.

Scheme of implementation of the switch manufacturing method is presented in Figure 9.

First on cylinder 15 of D diameter a figured slot of d width and h depth is made on external side. Then bushing 17 with radial hole 18 with a diameter not less than width of the slot $(d_1 \ge d)$ is pulled on the cylinder, and the bushing is welded to the cylinder. Then a concentric hole of D_1 diameter that is bigger than internal diameter of the slot $(h_1 < h)$ is made in the cylinder 15, and in this way slot 16 opens into cylindrical cavity of the hole. After this shaft 9 with a threaded seat for pin 10 is inserted into the hole, the seat is set opposite hole 18 in bushing 17 and the pin is screwed into the shaft outside the limits of internal diameter of the bushing, making in this way an element, which interacts with the slot, is located inside the cam, and does not exceed its dimensions in radial direction.

The switch (see Figure 8) operates as follows. In assembled form depending upon position of shaft 9

pins 10 may be arranged both in annular straight part of slot 17 and on the bend. Annular phase is a passive one; when the pin enters into it, coupling 11 is remote from a respective gear and its fingers 13 do not enter into hole 12 of the coupling. Instead of fingers and holes other designs of the couplings may be used. After turning of shaft 9 pin 10, which gets into the bend, moves hub with coupling 3 along shaft 9, simultaneously moving coupling 11 along shaft 5. Due to the fact that all gears 8 rotate relative couplings 11, fingers 13 get into holes 12 of the coupling 11, which moves to the fingers. Due to the fact that coupling 11 is connected with shaft 5 by key 14, rotation of gear 8 is transmitted over it to shaft 5. Reduction ratio and speed of rotation of initial shaft 5 depend upon the engaged coupling. As it is shown in Figure 8, the slots may be made in such way that during switching over not a single coupling will be engaged till the working one is withdrawn from the engagement. Intermediate positions correspond to idling of the engine and stop of the outlet shaft.

Such design is rather compact and allows, as it is seen from Figure 8, making gear boxes with a switch within overall dimensions of the gear cluster. In Figure 10 a welding manipulator is shown, which is made with such gear box according to described above example of turning inside of the switch with a cylindrical cam with a control element out. Such manipulator of 60 kg load capacity also weighs 60 kg.



NEWS

METHODS, EQUIPMENT AND TECHNOLOGIES OF PLASMA AND THERMAL-JET WELDING OF LIVE SOFT TISSUES AND THEIR APPLICATION IN SURGERY

Specialized equipment was designed and pre-clinical researches were carried out by employees of the E.O. Paton Electric Welding Institute, NASU, Ukraine, in collaboration with specialists of the A.A. Shalimov Institute of Surgery and Transplantology, AMS, Ukraine, on evaluation of applicability of the technologies of thermal-jet welding of live soft tissues.

This technology was designed in the frame of a new branch of medical science ---- hyper-thermal surgery ---- and is an organic continuation of live soft tissue welding and cutting method developed earlier. Taking into account the high price of argon-plasma equipment for live soft tissues welding, the necessity of designing other, more cost-effective equipment appeared. Its low cost, versatility and enhanced functional possibilities are the distinguishing features of thermal-jet technology, as well as equipment portability and self-sufficiency in application. The technology is characterized by a special mode of protein structure change without charring or complete destruction in tissues. A set of working tools, power unit and control panel with microprocessor are included into the equipment.

The working tool is completed with outlet port nozzles of different diameter and configuration. Nozzle diameter and configuration, as well as volume-temperature characteristics of gas flow, are determined by medical indications when carrying out concrete manipulations.

Minimum dimensions of working tools are from $200 \times 30 \times 30$ mm with the weight starting from 60 g.



Gas flow temperature is regulated in the range of 80–150 °C at heat flow from 900 up to 3335 W/m^2 at constant or pulse mode of feeding. Power consumption from the mains is up to 60 V·A.

Power block is made to A class of safety and enables operation from the mains of 220 V, 50–60 Hz or from self-sufficient DC power sources of 12 V (accumulator, car generator). Power block dimensions are 160 × $\times 205 \times 75$ mm. Operating time from 220 V mains is not limited. Operating time at operation from self-sufficient DC power source is limited by the source capacity.

NKMZ STARTED MANUFACTURING OF THE LARGEST MILL IN CIS

Novokramatorsk machine-building plant (NKMZ Company) (Kramatorsk, Donetsk region) started manufacturing the biggest ore crusher mill in CIS with drum diameter more than 9 m. It is designed for Lebedinsk MCW ---- a well-known in Russia enterprise on mining and concentration of iron ore and manufacturing of high-quality metallurgical raw materials. The machine will be manufactured and delivered to the customer already next spring, and it will be used for grinding ferrous metal ores.

New machine belongs to the type of wet autogenous grinding mills. Mill reliability and performance improvement will be achieved due to increase of loadcarrying capacity of the main bearings, modern pivot fastening and to other progressive technical solutions. New NKMZ mills are cost-effective for the customers also because they can be placed on old foundations.





KTM COMPANY AUTOMATES TIG WELDING OF MOTORCYCLE FRAME PARTS

One of the latest and successful KTM models is road motorcycle RC8. Its welds should not only be solid but should also have a nice appearance. That is why all RC8 frame joints are done by TIG welding. Traditional rippled weld is of high quality in TIG welding. However, there is one disadvantage: comparatively low productivity, it is by two thirds lower than that, which is obtained using more modern MAG welding processes.



Robot welding of fillet and V-shape welds with one bevel of two edges and gap of 2 mm, using MagicWave 4000 system

KTM technical experts dedicated three years of work to TIG process improvement, which resulted in 90 % automation of frame manufacturing. At this TIG welding efficiency is substantially raised in comparison with MAG process. The plant with robotic welding, where RC8 frames are welded, operates at full capacity in three shifts from the beginning of 2008.

Success was achieved particularly due to carrying out welds of a thin cross-section with good formation. Sheets of 1.2 mm thickness and pipes manufactured from high strength 25CrMo4 steel are used for welding. MagicWave of «Fronius» company is used as welding system.



Automated TIG welding with cold wire feeding in manufacturing of RC8 motorcycle frame

Specially designed system for pallet delivery performs the task of blank delivery to ABB robot until the frame is ready. Calculated thickness of fillet and V-groove welds with one bevel of two edges is 2 mm, and gaps that are required to be overlapped have 2 mm of width. «The heart and the soul» of a racing vehicle that consists of joint tubular sections can stand mechanical and dynamic stress when riding on the road and on a car racing route.

The reason for successful application of TIG welding is that it provides ideal (without reinforcement) shape of the front face of weld surface.

«Pulses» typical for TIG mode force filler material to melt in the shape of a series of teeth.

The use of a rather large, but strictly regulated heat input in TIG welding allows filling the gaps and obtaining an ideal penetration of weld root. Fusion at the beginning and at the end of the weld is considerably improved especially in pipe welding.

DEMAGNETIZATION CONTROL UNIT

Manufacturing a range of demagnetizing devices of LABS-7 series is developed by machine-building plant «VPERED» (St.-Petersburg, RF).



It permits performing pipe demagnetization and magnetic field compensation in the butt zone before welding with the aim of excluding «magnetic blow» effect and obtaining a sound welded joint. LABS-7 block is designed basically for magnetic field compensation in the welding zone that is the most effective method in assembling magnetized pipes and pipeline elements. Complementary circuit can be added by customer wish to this block set for increasing the power as well as LABS-7EM system for magnetic field localization. The unit can be used in small construction-mounting organizations.

Simplicity and convenience in operation is one of LABS-7K advantages.

Automated process of demagnetization control does not require special personnel training; control elements are reduced to minimum.

Minimum overall dimensions and mass, as well as simplified assembling schematic, require involvement of not more than one specialist that allows optimizing production process and raising labour productivity.



BRIEF INFORMATION

It is possible to shorten butt preparation process for welding from 3 min to 10 s, additionally buying LABS-7EM magnetic field localization system.

Demagnetizing block LABS-7K is also good for demagnetizing large-sized parts such as rotors and driving wheels of gas turbines, large-scale parts of complicated geometrical shape, ships and sub-marine bodies.

Due to application of special polymer-caoutchouc body coating and system of compensator coil induction heating, the unit can be used under any weather condition (even in the Far North conditions at ambient temperature of less than -40 °C).

The main parameters of the unit are supply voltage: 220--240 V, 50 Hz; power consumption: 2--5 kW; compensating field on the pipe: 200--1440 mm, 70--110 mT; adjustment accuracy: 0.1 mT; conditions: demagnetizing/compensation; fully loaded weight (with compensator coil): not more than 25 kg; preparation time: not more than 3 min; working time in compensation mode is limited; compensator coil: universal for 200--1440 mm diameter pipes; body: dielectric with humidity-resistant polymer coating.

COLOURLESS ANTICOR FOR WELDS AND METAL

CJSC «Elmid-Techno» (Moscow) offers «Iskra-3» agent in aerosol packing for reliable protection of welds and metal surfaces from corrosion.

It can be sprayed on metal surfaces without their preliminary preparation (directly on rust), as well as on coatings, used for weld section protection from molten metal spatter sticking (for example, sprayed on by «Iskra-2» agent).

Features of agent «Iskra-3» application: treated surface should be dry; flat, transparent, fast drying anti-corrosion coating, resistant against atmosphere influence, rubbing and impacts, is formed on the metal surface after its application; weld or other metal surface can be painted directly over coating, if required; protective coating can be removed from the surface by # 646 dissolver.



AS-150 INVERTER SOURCE

Manufacturing of modern professional AS-150 inverter power source for manual arc welding (MMA) of different metal structures under production and household conditions by all types of coated electrodes (except by electrodes for aluminium alloy welding) was established by Kiev enterprise OJSC «Artyom-Kontakt».

The advantages of the unit are determined by its capabilities:

- smooth welding current regulation;
- availability of «hot start» mode;
- availability of «anti-sticking» function;

• availability of open-circuit voltage decrease function;

• output parameters stabilization at mains voltage variation;

• availability of quick disconnect and safe current plugs;

• possibility of operation from low-powered electric mains;



• possibility of operation from self-sufficient power sources of 220 V, 50–1000 Hz, as well as from direct voltage 270–340 V;

• possibility of operation under the conditions of high dustiness.







The E.O. Paton Electric Welding Institute of

V.Yu. Belous (PWI) defended on 11th of June 2008 candidate's thesis on topic «Control of weld formation in tungsten electrode narrow gap welding (NGW) of titanium alloys with magnetically controlled arc».

The work is devoted to investigation of peculiarities of the processes that proceed in narrow gap welding with superposition of the external control magnetic field, which is used for controlling deviation of the welding arc and regulated melting of vertical wall of the gap.

Idea about mechanism of the welded joint formation in the narrow gap under conditions of the external magnetic field action was formulated. Influence of the control magnetic field on movement of the arc in the narrow gap and formation of a weld was determined. Regularities of the anode spot movement in the gap under action of external control magnetic field were established, and it was shown that value of its shift on vertical walls is directly proportional to value of cross component of the magnetic field induction and inversely proportional to the welding current. Gasdynamic pressure of the arc in NGW was evaluated using experimental-calculation method. It was shown that in NGW with a control magnetic field, in contrast to welding without external magnetic action, occur cross fluctuations of the weld pool metal, whereby main role in excitation of the metal fluctuations belongs to the arc pressure force. It was established that formation of the weld of optimal shape with equal over height penetration of vertical walls and absence of lacks of penetration and fusion are ensured at induction of the control magnetic field 8--10 mT. Working range of the magnetic field frequency reversing was established proceeding from conditions of prevention of lacks of fusion with surface of the previous layer and the base metal. Numeric value of factor of proportionality between shift of tungsten electrode from axis of the groove and maximum pulsation of the arc voltage was determined. It was established that tungsten electrode for NGW with external control magnetic field should have special shape of the tip. It was established that in NGW with an external reversible magnetic field occurs phenomenon of fluctuation of the non-cooled tungsten electrode tip, connected with action of the force from the external reversible magnetic field.

THESES FOR A SCIENTIFIC DEGREE

As a result of the investigations technological recommendations for NGW of structures from the PT-3V and VT23 titanium alloys of 20--100 mm thickness with application of a removable water-cooled forming backing were developed, which ensure absence of such defects of welded joints as lacks of penetration of vertical side walls of the gap and lacks of fusion of the weld metal with the base metal, and guarantee high level of mechanical properties of welded joints.



The E.O. Paton Electric Welding Institute of NASU.

I.A. Pributko (Chernigov State University of Technology) defended on 11th of June 2008 candidate's thesis on topic «Development of methods of evaluation and reduction of residual stresses in pressure sensors».

The thesis is devoted to search of ways for reduction of the level of internal stresses in metal-glass-silicon units of sensors of non-electric parameters for the purpose of increasing their operation capacities. Using developed on basis of the method of finite elements methodology of calculation of stress-strain state of multicomponent units of pressure sensors, which allows modeling stress-strain state in braze-welded structures of various configuration at the stage of manufacturing and operation, numerical experiments were carried out and main regularities of distribution of stresses in braze-welded multicomponent units of pressure sensors, depending upon geometric parameters of joints, physical-mechanical properties, and technology of manufacturing of the units, were established. It was found out on basis of established regu-

NASU.



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larities that reduction of level of maximum tensile stresses in the silicon membrane and the insulator is achieved by variation of geometry of the metal-glasssilicon joints.

Analysis of stressed state in the joints and mechanical tests of the glass-silicon units of semiconductor pressure sensors showed that for ensuring assigned metrological characteristics of the converters it is necessary to design unit of the sensor in such way that thickness of the sensitive element and thickness and height of the glass insulator be minimum, and thickness of the intermediate glass element and metal housing be maximum. Character of the applied load and presence of geometrical non-uniformity in zone of welded joints in the form of defects of a crack-like type may be hazardous from the viewpoint of strength of welded glass-silicon joints during tear work of the silicon membrane.

Using approaches of fracture mechanics of solid bodies with cracks, influence of defects, connected

with imperfect processing of glass parts for welding, was investigated. It was shown that for ensuring low probability of crack propagation in the pressure sensor unit depth of a defect of a crack-like form in the zone of a tubular glass part joint with a silicon membrane should not exceed certain dimensions according to conditions of brittle fracture. Results of experimental investigations using polarization-optical method confirm general conclusions concerning character of influence of geometry of a welded glass-silicon joint of a pressure sensor unit. Quantitative assessment of level of maximum stresses near joint zone of glass with a silicon membrane showed good correlation of results of experimental investigations of stresses by color-nomographic charts with the calculated results: differences in results constituted 12--18 %. On basis of the carried out investigations of stress-strain state of braze-welded multicomponent units of sensor equipment main principles of rational design of pressure sensor units were suggested.

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