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CONTENTS

45 YEARS OF WELDING ENGINEERING CHAIR

Dragan S.V., Kvasnitsky V.V., Romanchuk N.P., Solonichenko Yu.V. and Goloborodko Zh.G. Welding and cutting processes in shipbuilding of Ukraine
Vereshchago E.N., Kvasnitsky V.F., Romanovsky G.F. and Prosyanov O.F. Development of a highly dynamic machine for thermal cutting
Serbin S.I., Kvasnitsky V.V., Goloborodko Zh.G., Matvienko M.V. and Buryakov A.V. Modeling of gas phase composition in plasma cutting of ship hull steels
Kvasnitsky V.F. and Markashova L.I. Intensification of the process of diffusion bonding of heat-resistant alloys
Ermolaev G.V., Labartkava A.V. and Khiznichenko Yu.N. Allowance for plastic strains in computer modeling of stress fields in brazed joints made from dissimilar materials
Lebedev Yu.M. and Martynenko V.A. Application of a nonconsumable carbon-electrode arc at surface treatment of steels with a low hardenability

SCIENTIFIC AND TECHNICAL

Lobanov L.M., Poznyakov V.D. and Mikhoduj O.L. Effect of residual stresses on technological strength of welded joints of high-strength steel 14KhGN2MDAFB	29
Yushchenko K.A., Savchenko V.S., Chervyakov N.O. and Zvyagintseva A.V. Character of formation of hot cracks in welding cast heat-resistant nickel alloys	34
Savitsky A.M., Savitsky M.M. and Novikova D.P. Influence of welding speed and duration of periodical cooling on structure formation in welded joints of hardening steels in arc welding with thermal cycling	39
Bondarev A.A., Ternovoj E.G., Shvets V.I., Nazarenko S.V., Rassamakhin B.M. and Tarasov G.V. Structure and properties of thin-sheet EB welded joints in iron–nickel alloy 32NKA	44
BRIEF INFORMATION	

Pentegov I.V. Generalization of formula of K.K. Khrenov for	
determination of temperature of welding arc plasma	48
Zhernosekov A.M. Effect of electrode stickout on weld	
parameters in pulsed-arc welding of steels	51

Dear colleagues in welding: professors, teachers and associates of the Welding Engineering Chair of the Admiral Makarov National Shipbuilding University!

On behalf of scientists and associates of the E.O. Paton Electric Welding Institute and the Editorial Board of the «Avtomaticheskaya Svarka» Journal, we are congratulating you with the 45th anniversary of the Welding Engineering Chair.

Formed in 1959 in Nikolaev, the city of shipbuilders, the Chair has always coped with the tasks of raising the level of education of welding engineers, as well as development and implementation of high-efficiency technologies not only for shipbuilding, but also for almost all other engineering sectors. During the time of its existence, the Chair has educated about 3 thousand certified engineers, bachelors and masters for the welding industry of Ukraine, as well as the «near- and far-foreign» countries.

Topical scientific areas were formed and branch research laboratories and affiliated chairs were established at the major shipyards and marine engineering plants owing to the stirring creative activity of the team of the Chair headed during its early years by assistant professors A.I. Safonov and I.I. Dzhevaga, who combined scientific and industrial interests, and during the last 25 years by professor V.F. Kvasnitsky.

The Chair collaborates to advantage in training of specialists and R&D activity with the leading research and training centres of Ukraine, Russia, Germany and China, and takes an active part in certification of welding products in the south of Ukraine.

Scientists and specialists of the E.O. Paton Electric Welding Institute and the Editorial Board of the «Avtomaticheskaya Svarka» Journal are wishing happiness, wellbeing and success in their creative efforts to the staff, graduates and students of the Chair.

Below we are publishing related articles prepared by scientists, teachers and specialists of the Chair, covering results of their recent research.

WELDING AND CUTTING PROCESSES IN SHIPBUILDING OF UKRAINE

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Welding technologies used for the fabrication of ship structures at the shipyards of Ukraine are reviewed. Prospects for improvement of assembly-welding processes in shipbuilding are outlined.

Keywords: shipbuilding, technologies, welding consumables, equipment, labour consumption, production

At the beginning of the 1990s, the shipbuilding industry of Ukraine comprised 8 large shipyards, 8 machine-building plants and associations involved in the fabrication of marine equipment, 16 instrument-making enterprises, 30 research institutes and design bureaus, 40 subsidiary companies and factories to support the main production with a total staff of 250,000 people in 15 cities of Ukraine. The south region, being the major shipbuilding region in the former USSR, exported more than 100 ships with a dead-weight of 2.5 mln t to 20 countries throughout the world. The share of Ukraine in total output of shipbuilding of

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the USSR was 25.7 %, whereas the share of the northwest region was 23 and that of the St.-Petersburg region ---- 18 % [1].

Welding technologies and equipment applied at the shipyards were at a level of the world standards in science and technology. A great experience was accumulated in welding production management, using mechanised production lines and robotic systems for the fabrication of different types of ship structures.

Disintegration of the Soviet Union exerted a dramatically negative effect on the majority of the shipyards of Ukraine. Disrupted economic links, drain of a highly skilled manpower, physical and moral depreciation of fixed assets ---- all these factors led to destabilisation of the industry, loss of the accumulated experience and decrease in competitiveness of shipbuilding products in the world market. However, despite a difficult economical situation in the majority of the shipyards, a stable trend has been formed in the last years to recovery of production and growth of the amount of orders for national and foreign customers.

The Open Joint Stock Company «Damen Shipyards OCEAN» is progressing most successfully in this sector. It has already reached its full production capacity and builds ships on a turnkey basis. Its output during the last four years has grown more than 6 times. At present the main line of business activity of the Company is realisation of the shipbuilding program under 27 contracts, including construction of hulls and complete ships.

Restructuring of the state enterprise «Chernomorsky Sudostroitelny Zavod» (Black Sea Shipyard) has been completed. The Shipyard «Zaliv» has resumed its business activity, and now it builds for export a series of chemical carriers. The Sevastopol Marine Plant is involved in repair of ships.

Active shipyards and marine engineering plants apply now a large amount of welding operations. Welding technologies at the yards and plants are used at blanking shops, for prefabrication of units and sections, assembly of ship hulls on site and fabrication of marine engineering parts.

Ship hull components are made by the gas-oxygen and plasma cutting methods using CNC machines. At the same time, enterprises that failed to upgrade their production base and technologies experience a hot problem associated with ensuring precision of cut components to eliminate fitting operations in assembly of units and sections.

Less stringent assembly tolerances used at such companies, compared with those applied in machinebuilding, lead to a substantial increase in labour content of assembly operations. For example, in cutting of 8 mm thick parts and with a mean welding gap tolerance of 2 mm, the number of parts that do not fit the tolerance constitutes about 47 %. So, the labour content of fitting-blanking operations amounts to 24 % of that of the entire assembly process. Transition to a wider application of automatic and robotic welding requires much more stringent tolerances for assembly gaps or use of devices to compensate for errors in assembly of joints for welding. One of the ways of addressing this problem is to develop and implement integrated computer-aided systems for design and cutting of hull structure components. The use of such a system and the NUMOREX plasma cutting machines at «Damen Shipyards OCEAN» provided the required precision in the manufacture of units and sections and allowed avoidance of fitting operations at the stage of staple assembly of the hulls.

However, the experience of operation with the NUMOREX machines showed a relatively low coefficient of utilisation of the machine time (about 60%), which is associated with a traditionally applied technology of location and spacing of sheets and removal of finished parts from a shearing frame. Application of the TELEREX TXB-10200 machine comprising four shearing frames installed in water pools made it possible not only to cut symmetric parts of the left and right boards simultaneously and, if necessary, with an edge bevel, but also to lay out the parts, excluding almost any downtime of the machine.

The Kherson Shipyard successfully uses air-plasma cutting with an addition of water to plasma, which also decreases distortions of the cut parts. Advantages of this method include the use of oxygen as a plasma gas, as well as cathodes resistant to oxygen-containing gas mixtures.

Welding of ship structures is performed by the automatic submerged-arc and mechanised gasshielded methods. Automatic submerged-arc welding is used for the fabrication of panels of sections and joining of principal frames, while mechanised welding is used for the fabrication of units and sections, as well as assembly of the ship hulls. Gas-shielded welding outgrew into automatic welding of assembly joints of the ship hulls and robotic welding of high frame intersection joints in the fabrication of sections.

Experimental application of the IGM robotic gantry system at «Damen Shipyards OCEAN» showed its high efficiency under conditions of high-quality assembly operations. Deviation of the fillet weld line should not exceed ± 2 mm, while the in-process adaptation of the welding torch to arc parameters is required at higher deviations. Use of the robotic gantry systems comprising two or more welding robots for the manufacture of large sections with frames more than 1.5 m high provides rise in productivity of welding processes and reduction of working hours in the fabrication of ship structures. The positive experience has been gained in the world practice with the application of such integrated welding systems [2].

Application of laser and hybrid welding is a promising trend in upgrading of welding technologies used in domestic ship hull construction. It is expedient to use laser welding for the fabrication of panel structures, on which the increased requirements for accuracy and appearance are imposed. This applies, for



example, to ship superstructures and deck houses [3]. Hybrid welding (laser + MIG) provides deep penetration that ensures the laser welding process and a good quality of welding over the gap, which is required for the majority of the ship hull structures [4].

Satisfactory appearance of ship structures is achieved in current industrial practice by using thermal dressing, which is performed with the multi-nozzle gas torches at a temperature of 350 °C, which is the case of «Damen Shipyards OCEAN».

Orientation of shipbuilding of Ukraine to performing export orders led to the necessity to meet more stringent requirements to surface preparation of welds for painting. Companies that do painting of ships and guarantee high quality of paint work to ship-owners require an appropriate surface preparation of welded structures. But, as shown by experience, providing the required quality of the surface involves high material and labour expenditures. These expenditures can be reduced by using mechanised gas-shielded solid or flux-cored wire welding. In the EU countries, welding in argon-based gas mixtures and CO₂ welding using solid wires are dominant among the mechanised welding methods [5]. CO₂ welding traditionally used in shipbuilding is replaced to advantage by welding in gas mixtures. Electric arc welding in a gas mixture consisting of 80 % argon and 20 % CO₂ using wires Sv-08G2S and Sv-10GSNT is much superior in technological characteristics to CO₂ welding. It provides the required quality of the weld surface without extra machining [6]. However, expansion of the application of welding in gas mixtures needs further development of domestic welding consumables and equipment, as well as arrangement of centralised systems to provide assembly-welding shops with gas mixtures.

A reserve to improve quality of welded joints and decrease labour content of the fabrication of ship structures due to avoidance of postweld machining operations lies in expanding the scopes of application of flux-cored wire welding.

Our country has accumulated experience in the development, manufacture and application of fluxcored wires (PP). A wide range of this type of wires was developed with a direct participation of the E.O. Paton Electric Welding Institute [7]. However, there are some factors that limit application of the flux-cored wire welding technology at domestic enterprises. Along with a high cost of this wire, they include also the absence of, for example, home-made feeding mechanisms and hose-type holders. Despite this fact, the integrated approach to estimation of cost effectiveness, allowing for the costs of not only materials and consumables, but also the costs of post-weld operations, proves profitability of using flux-cored wire welding in shipbuilding [8].

Today the Ukrainian market of welding consumables offers a sufficiently wide range of home- and foreign-made flux-cored wires. Among them are products of such known foreign companies as Lincoln Electric, Hobart, ESAB, S.A.F.-Oerlikon, Thyssen-Bohler, Kobeko, Elge, etc., as well as products of a number of Ukrainian manufacturers. Worthy of special notice in this respect is the «Dneprometiz» Company. Rutile wires PP-AN59 and PP-AN63 were developed for welding carbon and low-alloy steels. These wires allow welding in any spatial position.

Wires PP-AN70 and PP-AN72 with a metal core were developed to increase yield of the deposited metal to 95 %. These wires can be used for CO_2 welding and welding in gas mixtures [5, 9].

The Kherson Shipyard uses home-made flux-cored wire PPs-TMV7 with a diameter of 1.2 mm, and «Damen Shipyards OCEAN» ---- imported thin flux-cored wires PZ 6113S, PZ 6111, OK Tubrod 15.13S and OK Tubrod 15.14 for mechanised CO_2 welding. At the Japanese shipyards the mechanised and robotic gas-shielded welding processes are performed using flux-cored wire with a diameter of 1.2–1.6 mm [10].

Increase in volumes of application of flux-cored wires, which are not inferior to the imported ones, but, at the same time, are less expensive, requires an urgent certification of the Ukrainian manufacturers of welding consumables and equipment by international certification societies.

An important problem of modern shipbuilding is to decrease labour consumption and improve labour conditions in performing welding operations under staple conditions and joining panels and frames under assembly conditions.

One-sided welding using flexible self-adhesive backings is a promising method for welding such joints. It allows the process productivity to be raised, a welding operator to work outside the discomfortable working zone and quality of a welded joint to be improved.

The shipyards of Ukraine use wire PP-AN25 for welding in a horizontal plane with semi-forced weld formation using ceramic backings to form the weld root, and wire PP-AN39 ---- to fill the groove in weld formation on a water-cooled shoe.

A large amount of welding operations using systems with a different-diameter pipes is applied in shipbuilding. The high quality of treatment of parts and welding of joints in pipe production provides operation of modern cutting and welding equipment with programming of the treatment mode and tool movement line. Experience of operation with the TUBOSES and POLYSOUD machines at the shipyards of Ukraine showed their high efficiency under conditions of availability of highly skilled engineering and operating personnel.

The marine engineering industry is characterised by a continuous improvement of structural materials. Heat-resistant superalloys, composite and non-metallic materials used in ship gas turbine construction require the use of such special technologies as electron beam welding, brazing, electron beam evaporation and ion plasma sputtering, as well as laser treatment, plasma and gas circulation spraying. In particular, the Nikolaev Gas Turbine Construction Works «Zo-

rya-Mashproekt» successfully uses vacuum technologies developed as far back as the 1980s--1990s, which are still classed with the advanced welding technologies.

The near-term perspective of Ukraine to join the international trade community will provide a strong base for increasing the demand for ships. According to estimates made by analysts, the world-wide amount of shipbuilding orders will be not less than 200 mln register tons. In this connection, one of the problems of shipbuilding will be upgrading of welding production through a wide and prompt application of advanced technologies, as well as promising welding methods and equipment, which without doubt should lead to full utilisation of the available capacities and growth of the Ukrainian shipbuilding industry as a whole.

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DEVELOPMENT OF A HIGHLY DYNAMIC MACHINE FOR THERMAL CUTTING

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A new approach is considered to construction of the systems of control of electric drives in thermal cutting machines, proceeding from the condition of implementation of the assigned motion trajectories, aimed at expansion of the sphere of their application and improving the cost-effectiveness of their use.

Keywords: thermal cutting, computer control, feed electric drive, automatic following, positioning, cutting unit

Current blanking production features an extensive application of machines, equipment and technologies of thermal cutting [1, 2], providing an effective solution of the problem of automation and mechanization of the production process simultaneously with improvement of the efficiency and provision of the required accuracy and quality of the cut out blanks (specified by GOST 14792--80 and similar standards of Germany, France and other countries). New foreign-made machines and equipment for thermal cutting are sophisticated and expensive. Therefore, there arose the need to develop less expensive, but reliably operating thermal cutting machines (TCM).

The purpose of this work is development of a highly dynamic machine for plasma and laser cutting. One of such machines was developed by Admiral Makarov National Shipbuilding University together with NPP «Ukrtermmash». KRISTALL-TM machines have been put into production, they are supplied to a number of enterprises, where they operate successfully.

The above machines, pertaining to industrial welding robots of IF class [3], include the following elements: cutting machines, control systems and propulsion systems, consisting of actuating drives and mechanisms. In order to ensure a high accuracy of cutting out the parts, they should be regarded as one integrated complex.



Figure 1. Basic model of KRISTALL-TM machine

In terms of accuracy, quality, effectiveness and efficiency of severing various materials, the most rational of the currently available technologies are plasma (including HiFocus, HiDefinition, HiSpeed, LongLife and other technologies) and laser cutting technologies [3]. Accuracy of plasma cutting is up to 0.25 mm at repeatability of ± 0.175 mm, and that of laser cutting is 0.10 and ± 0.05 mm, respectively [3]. However, high speeds of plasma and laser cutting are required [3, 4]. Cutting speed in the typical machines is close to 12 m/min, and that of accelerated motion to 20 m/min. It is obvious that these requirements can only be implemented in the modern highly dynamic and high speed machines.

Experience of operation of locally made machines manufactured in the past years showed that their transportation-mechanical equipment is rational and durable [1, 2]. Therefore, design of the mechanical part of the new machine is based on the typical most rational and well-established structure of gantry type with a rectangular lay-out [1, 2]. A machine for cutting materials with a two-side precision rack-and-pinion drive is shown in Figure 1. It is a mechanical system, consisting of sufficiently rigid assemblies, connected by elasto-dissipative ties.

Clearance in the rack-and-pinion is selected automatically, using force pressing of the gear to the rack, and in the gearing of reduction gears by adjustment of the interaxial distances. Guides of the travel carriage include self-regulated side rollers, preventing the frame sliding down. In this case, the accuracy of the mechanical part is increased by an order of magnitude.

A successful solution of the problem of developing machines with modern cutting tooling, reduction of the time and material costs for technology design, simplification of TCM operation and repair have been achieved due to modular design of the machines, namely design, allowing for international standardization and unification of typical components and modules, both as regards their mechanical components and information-control and propulsion systems. All this permitted implementing the idea of easily readjustable versatile TCM, which created the prereq-

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uisites for robotization of small-batch and batch production under the conditions of small and mediumsized businesses.

Important characteristics of the cutting machines are the accuracy of positioning and motion repetition. In keeping with the definitions, introduced by the Japanese experts [5], TCM following systems belong to semi-closed systems, where indirect measurement of the elastic mechanism position is performed. Accuracy of machine positioning in this case is almost completely determined by the error of the mechanical system. In this case, the high accuracy of repetition and positioning can be provided by the respective correction in the PNC device (measuring the clearances and errors of transmission). For instance, a clearance in a structure semi-closed by its position is compensated in keeping with the following expressions:

$$\begin{split} N_{\mathrm{ass}\ \phi} &= N_{\mathrm{ass}\ \phi}^0 + N_b \ \mathrm{at} \ N_{\omega} > \mathbf{0}; \\ N_{\mathrm{ass}\ \phi} &= N_{\mathrm{ass}\ \phi}^0 - N_b \ \mathrm{at} \ N_{\omega} < \mathbf{0}, \end{split}$$

where $N_{ass \phi}^0$ is the code of assigned displacement without allowing for the clearance; N_b is the code, which is the numerical equivalent of the pre-measured clearance.

Analysis showed that in order to achieve the required accuracy and cleanliness of the surface of the item, processed in TCM, the main problem is to achieve uniform motion of feed mechanisms at a fast response of the regulation systems, i.e. satisfy the requirements made of an automated electric drive. Solution of this problem requires a new approach to designing the following systems of control by the trajectory.

For the considered case, the machine performance and quality of the products made by these machines are largely dependent on the quality parameters of electric drive dynamics. Significant factors determining the dynamic properties of electric drives are instability of their parameters and external impacts, presence of non-linearity and elastic links, which does not permit fully using the potential of transistorized converters. The accuracy of reproducing the blank contour depends on the part shape and speed of motion around the contour. TCM motion at corner tracing is accompanied by dynamic effects. Mechanical systems of drive--gantry and drive--support are subjected to a pulsed impact. Pulse duration is determined by the following expression:

$$t_{\rm pulse} = \pi R / (2v),$$

where R and v are the radius and speed of corner tracing, respectively.

For gantry robots for plasma and laser cutting the radius of the circumference, approximating the right angle, is equal to 1--4 and 0.1--1 mm, respectively. As the time of corner tracing is commensurate with the periods of natural oscillations of the gantry and

support (frequencies of natural oscillations were 8.58 and 21.32 Hz, respectively), it is anticipated that the shape of the initial angle will be distorted due to dynamic effects.

Let us consider this approach for the case of the systems of control of the position of an elastic dynamic object with subordinated adjustment of three variables, namely position, speed and current. More widely used systems now are those based on the principle of subordinated regulation [3]. TCM positioning mechanisms, as a rule, have a drive with DC motors with separate excitation. Therefore, let us restrict ourselves to consideration of just this drive.

Elasticity usually leads to appearance of mechanical oscillations of various kind. In this case, damping (compensation) of elastic oscillations due to an automated electric drive is reduced to the need to simultaneously lower the dynamic coefficients of amplification of the regulators of speed and current to such an extent that the resonance surges of amplitude frequency characteristic did not go into the prohibited region. At regulator readjustment the realizable pass band of position loop in principle remains to be much lower than the mechanical resonance frequencies, i.e. the Q-factor of the following system becomes significantly lower, and dynamic properties of the modern transistorized electric drive are not used sufficiently.

The high quality of regulation, suppression of elastic oscillations and normal functioning under the conditions of variation of drive parameters and tuning coefficients of subordinated regulation circuits in a broad range are achieved by correction of position loop of the actuator shaft due to introduction into the position regulator of an active rejector with independent control of zeros and pluses of its transfer function:

$$W(p) = \frac{b_0(p^2 + \omega_0^2)}{\omega_0^2(p^2 + b_1p + b_0)}$$

(here *p* is the operator of Laplace function; b_0 , b_1 are the adjustment coefficients; ω_0 is the frequency of filter rejection, which is selected equal to that of elastic oscillations) and of a new circuit of acceleration control, which is a means of elimination of indeterminacies in the object behaviour (Figure 2). It is obvious that for compensation of elastic oscillations, it is necessary to introduce into the filter transfer function a zero, equal to the value of natural oscillation frequency of a two-mass system.

The procedure of synthesis of the additional impact loop is conducted in two stages, namely first the acceleration is determined, at which the assigned trajectory is realized, then the corrective action is calculated.

For realization of the above solution it is necessary for the system dynamics by the output variable ϕ to correspond to equation



Figure 2. Functional sequence of a feed servo drive of TCM with PNC: PR, SR, CR — position, speed and current regulators, respectively; PS, SS — position and speed sensors, respectively; DCM — DC motor; RG — reduction gear; A — actuator; $\varphi_{ass}(I_{ass})$, $\varphi_{act}(I_{act})$ — assigned and actual value of the angle of rotation (position) of the actuating motor, respectively; I_{ass} , I_{act} — linear displacement of the actuator, assigned and actual, respectively; U_{PR} , U_{SR} , U_{CR} , U_{CS} — output voltage of position, speed, current regulators and current sensor

$$\overset{\cdot\cdot}{\phi}^{*} + a_{1} \overset{\cdot}{\phi}^{*} + a_{0} \phi^{*} = a_{0} \phi_{ass} \ (a_{0}, a_{1} > 0),$$

where $e = \ddot{\varphi}^* - \varphi^-$ is the program value of the second derivative of the controlled quantity, which is the master control; $\ddot{\varphi}^*$ determines the deviations of the actual behaviour from the sought e(t) = 0; a_0 , a_1 are the constant parameters, selection of which determines the shape of trajectory $\varphi^*(t)$; φ_{ass} is the assigned value of output variable (angle of rotation of the engine shaft).

In order to compensate for this mismatch, feedback is introduced using error-closing control

$$\mu(t) = ke(t),$$

where *k* is the coefficient of amplification, determined from the conditions of provision of the assigned error in the statics and dynamics. Program value of $\ddot{\phi}^*$ was established according to equation

$$\ddot{\phi}^* = a_0(\phi_{ass} - \phi) - a_1\dot{\phi}$$

Realization of control requires multiple differentiation of the observed φ signal. Obtaining accurate values of the derivatives is made difficult by random disturbances, observation noises and hardware noises. Considering that the motor moment is proportional to current in the armature circuit, the magnitude of the second derivative of the angle of rotation with the known motor parameters can be calculated with the accuracy down to the values of the moment of resistance, derived at measurement of armature current $i_{\rm ar}$. This means that $d^2\varphi/dt^2$ value can be evaluated, using a proportional link, if a directly measured armature current $i_{\rm ar}$ is applied to its input. Now let us determine the action signal

$$\mu(t) = k[a_0(\varphi_{ass} - \varphi) - a_1\dot{\varphi} - a_2i_{ar}]$$

and close the output of the additional loop to the input of a proportional-integral current regulator (see Figure 2). Use of additional feedback allows construction of a closed system with motions independent in the asymptotics on variation of object parameters in a broad range and for uncontrollable external disturbances. As a rule, during system design it turns out to be possible to assign the desirable regulation time t_r^* at execution of constant actions, as well as indicate the transition process characteristics, namely overcorrection σ , index of oscillation ξ , etc. These data allow determination of the transfer coefficients of the decision amplifier a_0 , a_1 . As was noted above, in the position loop it is necessary to satisfy a combination of contradictory requirements, namely a wide pass band (high following accuracy at program reproduction) in the contour mode and aperiodic characteristic in the mode of actuator positioning. Proceeding from the required dynamic properties, let us present in a standard form the reference model, where the order of differential equation is not higher than the second order:

$$\tau^2 \dot{\phi}^* + 2\xi \tau \dot{\phi}^* + \phi^*_{ass} = \phi_{ass}$$

where τ is the time constant; ξ is the coefficient of oscillation attenuation. Comparing the latter equation with the assigned one, we have

$$a_1 = 2\xi \tau^{-1}, \quad a_0 = \tau^{-2}.$$

It is known that, if $\xi = \sqrt{2}/2$ (optimum module adjustment), then duration of the transition process is $t_r^* \approx 3\tau$, and overcorrection is 4.3 %. These are the most preferable characteristics for feed drives. Assuming $\xi = \sqrt{2}/2$, we have

$$a_1 = \sqrt{2\tau}^{-1}; \quad a_0 = \tau^{-2}; \quad \tau = t_r^*/3.$$

Then, the required *k* value is calculated, at which the assigned motion trajectory is realized.

Experimental studies (Figure 3) of electric feed drives with loops of current, speed and position, having standard settings at 3- and 5-fold variation of object parameters (inertia masses of an elastic object, drive parameters), action of static and impact load, as well as at great misalignments of the subordinated control loops, bringing a typical system into a nonworking condition, showed that introduction of a rejector and acceleration control loop provides an ef-



Figure 3. Transition processes in the following feed system of TCM: 1 — with switched on correction with 50 rad/s Q-factor; 2 — without correction with Q-factor of 8 rad/s; ω_2 — actuator angular speed

fective suppression of resonance phenomena in feed drives (in absence of the latter, oscillations of the second mass with 8--10 Hz frequency are observed), greater uniformity of rotation (coefficient of non-uniformity is reduced by approximately 1.5 to 3 times), high quality of regulation (aperiodic transition process has an assigned shape with process time of 0.01 s and overcorrection of 5 %, which corresponds to the loop pass band of not less than 50 Hz). In this case the pass band of the following feed system of PNC robots becomes more than 5 times wider at program reproduction in the contour processing mode (from 10--15 up to 50 Hz and more), i.e. it overlaps the region of its resonance frequencies. Note that in the initial system the regulation is over in 0.15 to 0.20 s $(t_{\rm r} \approx 3(\omega_{\rm cf})^{-1})$, where $\omega_{\rm cf}$ is the cutoff frequency).

Application of new drives in TCM promoted oscillation suppression, also at simultaneous operation of both the drives at a varying speed, this providing a rather accurate reproduction of the linear and circular sections, sharp angles (the most complicated



Figure 4. KRISTALL PPLTS-2.5 machine after upgrading

element for following), high dynamic reaction (maximum deviation from the specified trajectory was just 0.02 mm), maximum motor speed achieved by the system (about 20 m/min).

At the same time, performance of labour-consuming adjustment is not required either at the stage of TCM commissioning or at their upgrading and repair.

An important part of the machine is its control system. TCM use a device of CNC type ---- modern analog of program numerical control. Programs are recorded in standard ESSI or ISO codes (GOST 20.999--83), and are a sequence of lines, where for each section the coordinates are assigned for the point, to which it is necessary to move, as well as the method and parameters of interpolation during performance of such a motion [6].

Approaches, described in this work, were used to develop a family of new highly dynamic cutting machines. The family includes 11 modifications adapted to the component layout features of serviced equipment, as well as the nature (batch-size) of manufacturing. Basic machine is a further development of TCM of KRISTALL type.

Developed machines envisage processing of sheets up to 3.6 m wide, up to 300 mm thick, depending on cutting fixtures. They can be effectively used in medium-sized and large businesses. For enterprises already having TCM with digital or photo-copying de-



Figure 5. GRANAT PPIKP-3.2 machine after upgrading

vices, a technique of their upgrading and fitting with additional fixtures has been optimized. After upgrading the machine efficiency rises by 30--50 %, their functional capabilities are expanded, and high cutting quality is guaranteed.

Specification of the basic TCM variant is given below:

Maximum width of the processed sheet, mm
at cutting.
oxvgen 300
plasma with piercing 528
plasma at cutting-in from the edge
laser 0.510
Speed of actuator displacement along axes x, y,
m/s (m/min):
rapid traverse
maximum working 0.2 (12)
Maximum averaging along axes x, y, m/s^2 2
Positioning error on 10 m length, mm $\dots \pm 0.2$
Repetition error, mm ±0.1
Range of reproducible conditions 1.10°

Machines made and upgraded (Figures 4 and 5) over the last years by «Ukrtermmash» have been successfully operating in shipbuilding and mechanical engineering, metallurgical and construction industry enterprises of Ukraine, Belarus and Russia [1, 2]

The above approach to construction of TCM of the new generation, based on the modern system of electric equipment not only meets the requirements of the international standards, but also ensures a strict performance of all the system tasks under the conditions of a significant parameter instability, presence of mechanical resonances and random variations of the load.

Developed control methods and algorithms have significantly widened the spheres of TCM application and ensured increase of economic effectiveness from their introduction. Approaches to improvement of electric drives and their realization, considered in this work, provide preservation of the required optimization characteristics during the entire period, «factory readiness» after machine repair and upgrading.

CONCLUSIONS

1. Application of the considered approach to dynamic synthesis of feed drive regulators of the new generation TCM significantly improved the response and reduced the positioning errors, this leading to widening of the sphere of application of thermal cutting machines and improving their economic effectiveness.

2. Developed basic model of an all-purpose thermal cutting machine is optimum in small-batch production and suitable, in particular, for laser cutting.

3. Modular design of the machines provides a fast response to customer requirements and promotes further improvement of production.

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MODELING OF GAS PHASE COMPOSITION IN PLASMA CUTTING OF SHIP HULL STEELS

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Two plasma-chemical systems, namely air low-temperature and air-vapour plasma, are considered. In keeping with the kinetic diagram, including 258 reactions, the design method was used to determine the composition of the gas phase in weight fractions of the components.

Keywords: air-plasma cutting, plasma-forming medium, air-vapour plasma, plasma components, saturation with nitrogen, pore formation

Air-plasma cutting is the basic method in manufacture of ship hull elements of 5--30 mm thickness. As to the efficiency, it is greatly superior to gas oxygen cutting, however, the increased susceptibility to pore formation in welds is observed in submerged arc welding. The air-plasma cutting at up to 14 mm thickness of metal to be cut has the highest effect on pore formation [1].

It was found that the main cause of pore formation in welding is nitrogen dissolved in cut edges and being in a state of an oversaturated solution [2--4]. It was stated experimentally that water adding to a plasmaforming air prevents the pore formation.

The aim of the present work was to study, using modeling, the gas phase composition in plasma cutting with water addition to the plasma-forming air.

It was established in work [5] that the temperature near the arc boundary is equal approximately to 5000 K. It follows from works [6, 7] that temperature near the arc boundary (from different calculation methods) equals from 3600 to 6000 K. Coming from these data the temperature of gas, reacting with metal, was taken equal to 5000 K, and the arc volume was constant that allows us to compare the processes proceeding in the arc in plasma cutting with processes in a plasma-chemical reactor [8].

A mathematical model of chemical transformations was developed from the conditions of the plasmachemical reactor. This model describes the chemical processes in a mixing chamber and a main zone of arc burning. Mathematical description of the processes is based on the fact that stirring occurs so intensively that the temperature and concentration of components in the reactor itself are equal to the temperature and concentration of products which are evolved from it, and the mass consumption of a working medium is constant here [9].

In the present work the composition of the gas phase was determined by calculations using a kinetic diagram which included 58 reactions with 58 possible components [10]. Two types of plasma-chemical systems were studied: air low-temperature and air-vapour

plasma, modeling the case of water supply to the plasmatron nozzle channel.

Equations of conservation of components and energy have a form [11, 12]

$$G(Y_k - Y_k^{\text{in}}) - \omega_k M_k V = 0; \qquad (1)$$

$$G\sum_{k=1}^{K} (Y_k h_k - Y_k^{\text{in}} h_k^{\text{in}}) - Q = 0, \qquad (2)$$

where G is the mass consumption of gas mixture component; Y_k is the mass share of component k at outlet; ω_k is the rate of formation of component k in volume unit as a result of chemical reaction; M_k is the molecular mass of k-th component; V is the rector volume; K is the total quantity of components; h_k is the enthalpy of elements; Q is the heat flow in reactor; index in corresponds to the conditions at inlet.

Conditional time of component duration in the reactor is defined from expression

$$\tau_{\rm r} = \rho V / G.$$

Density of mixture ρ is calculated from the equation of perfect gas state

$$\rho = pM/(RT)$$

where p is the mixture pressure; M is the molecular mass of mixture; R is the universal gas constant; T is the mixture temperature.

It was assumed in determination of a rate of the component formation that the coefficient of rate of an option elemental reaction is expressed by the Arrhenius' equation.

The elemental reactions considered can be presented in a general form

$$\sum_{k=1}^{k} V_{k,i} X_{k} = \sum_{k=1}^{k} V_{k,i} X_{k},$$

where i = 1, ..., I; $V'_{k, i}$, $V'_{k, i}$ is the stoichiometric coefficient, respectively, of the *k*-th initial component and *k*-th product in the *i*-th reaction; X_k is the molar fraction of the *k*-th component.

Rate of formation of the k-th component in the volume unit as a result of chemical reaction is

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11

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Figure 1. Composition of gas phase in air-plasma cutting

$$W_k = \sum_{i=1}^I V_{k,i} q_i,$$

where k = 1, ..., K; $V_{k, i} = V'_{k, i} - V'_{k, i}$ is the stoichiometric coefficient of the *k*-th reaction; q_i is the rate of proceeding of the *i*-th reaction, which is determined by equation

$$q_{i} = K_{f, i} \prod_{k=1}^{K} [X_{k}]^{V_{k,i}} - K_{r, i} \prod_{k=1}^{K} [X_{k}]^{V_{k,i}},$$

where $[X_k]$ is the molar concentration of the *k*-th element; $K_{f, i}$, $K_{r, i}$ are the constants of rate of proceeding of direct and reverse *i*-th reaction, respectively.

Constant of direct reaction rate is expressed by formula

$$K_{f,i} = A_i T^{B_i} \exp\left(-\frac{E_i}{RT}\right)$$

where A_i is the preexponential factor, mol/(cm K); B_i is the exponent of degree at temperature T, K; E_i is the energy of activation, cal/mol; R is the universal gas constant, cal/(mol K).

Constant of reverse reaction rate will be found by formula

$$K_{r, i} = \frac{K_{f, i}}{K_{c, i}},$$

where $K_{c, i}$ is the constant of equilibrium expressed by concentrations.

Value $K_{c, i}$ can be determined easier by thermodynamic properties by pressure

$$K_{c, i} = K_{p, i} \left(\frac{P_{\text{atm}}}{RT} \right)^{\sum_{k=1}^{K} V_{k, i}},$$

where $K_{p, i}$ is the constant of equilibrium, expressed by a partial pressure of components; P_{atm} is the atmospheric pressure equal to 100 kPa.

Constant of equilibrium $K_{p, i}$ will be obtained from expression

$$K_{p, i} = \exp\left(\frac{\Delta S_i}{R} - \frac{\Delta H_i}{RT}\right)$$

where the values Δ indicate the possible changes in transition of initial elements into products of reactions:

$$\Delta S_i^0 = \sum_{k=1}^K V_{k, i} \Delta S_k^0;$$
$$\Delta H_i^0 = \sum_{k=1}^K V_{k, i} \Delta H_k^0,$$

where ΔS_k^0 , ΔH_k^0 are the standard changes of enthropy and enthalpy of the *k*-th element, respectively.

Temperature and mass fractions of components are the solution of system k + 1 of non-linear algebraic equations (1) and (2).

First, the relationships of molar fractions of components of X system without water vapour additions were obtained (Figure 1). Chemical interaction in this case was regarded in the form of reactions:

$\widehat{\mathrm{I}}_{2} ightarrow 2 \widehat{\mathrm{I}}$;	$N + O \rightarrow NO;$
$2 \widehat{I} \ \rightarrow O_2;$	$N + 2O \rightarrow NOO;$
$2N \to N_2;$	$2N + O \rightarrow N_2O;$
$N_2 \rightarrow 2N;$	$N + 2O \rightarrow NO_2$.

With a rise in temperature from 1000 to 3000 K the molar fractions of nitrogen oxide NO and atomic oxygen are increased. Change in temperature from 3000 to 5000 K leads to an abrupt increase in content of N and NOO at thermal decomposition of molecular nitrogen. Content of compounds NO_2 and N_2O is negligible. Molar fractions of molecular and atomic oxygen at 5000 K temperature is equal almost to zero.

Modeling of behaviour of water entered the plasmatron nozzle is reduced to gas-phase reactions of the ionized air with a water vapour. Chemical interaction in the system, modeling the case of air-vapour plasma, can be presented in the form of the following reactions [8, 10]:

Í $_2 \rightarrow 2$ Í ;	$\begin{array}{l} \mathrm{H} + \mathrm{N} + \mathrm{2O} \rightarrow \\ \rightarrow \mathrm{HONO}; \end{array}$	$N + 2O \rightarrow NOO;$
$N_2 ightarrow 2N;$	$N + H \rightarrow NH;$	$2N + H \rightarrow NNH;$
$\widehat{I}_{_2} \rightarrow 2 \widehat{I}$;	$2 \acute{I} \ + 2 \widehat{I} \ \rightarrow \acute{I} \ _{_2} \widehat{I} \ _{_2};$	$\mathrm{N}+2\mathrm{H}\rightarrow\mathrm{NH_2};$
$2H \to H_2;$	$\mathrm{H}+\mathrm{2O}\rightarrow\mathrm{HO}_{\mathrm{2}}\text{;}$	$\mathrm{N}+3\mathrm{H}\rightarrow\mathrm{NH_3};$
$2N \rightarrow N_{2};$	$2 {\mbox{i}} \ + {\mbox{i}} \ \rightarrow {\mbox{i}} \ _{_2} {\mbox{i}} \ ;$	$\begin{array}{l} H + 3O + N \rightarrow \\ \rightarrow HOONO; \end{array}$
$2O \rightarrow O_2;$	$2N + O \rightarrow N_2O;$	$H + N + O \rightarrow HNO;$
$H + O \rightarrow OH;$	$N + 2O \rightarrow NO_2;$	$H + 2N + O \rightarrow HNNO.$
$N + O \rightarrow NO^{\circ}$		



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Figure 2. Change in degree of conversion of molecular nitrogen F_{N_z} in air plasma depending on molar fractions of water vapour: $1 - H_2O = 0$; 2 - 0.10; 3 - 0.17; 4 - 0.25; 5 - 0.35; 6 - 0.50

Thermodynamic properties of components are applicable up to 5000--6000 K temperatures [8, 10]. Dependencies of molar fractions of stable products and intermediate compounds on temperature under the air-plasma conditions were investigated at molar fractions of water vapour in mixtures with air plasma, equal to 0.10, 0.17, 0.25, 0.35 and 0.50.

Figure 2 shows the dependence of conversion of molecular nitrogen $F_{\rm N_2}$ on temperature at different additions of water vapour. Conversion degree is assumed to be the ratio

$$F_{\rm N_2} = 1 - \frac{X_{\rm N_2}}{X_{\rm N_2}^0},$$

where X_{N_2} is the equilibrium share of molecular nitrogen in plasma-chemical system; $X_{N_2}^0$ is the molar fraction of molecular nitrogen in the initial mixture which is supplied to the plasma cutter.

Thus, the degree of conversion is similar to the degree of thermal decomposition and nitrogen binding in a plasma jet. At conversion degree $X_{N_2} = 1$ all the molecular nitrogen is available in the form of bound compounds or in atomic form.

As is seen from Figure 2, the significant growth in degree of conversion in the plasma-chemical system with additions of water vapour is occurred with increase in temperature. Thus, at 5000 K temperature the degree of nitrogen conversion is 0.21, 0.31, 0.39, 0.41 and 0.66 at molar fraction of water vapour, respectively, 0.10, 0.17, 0.25, 0.35 and 0.50 (for comparison $F_{N_2} = 0.06$ for air plasma at mentioned temperature). It is important here that the molar fraction of the atomic nitrogen is decreased in the entire temperature interval at water adding (Figure 3). This proves that conversion of molecular nitrogen is proceeding with formation of not atomic nitrogen, but complex compounds being non-soluble in metal. According to calculations, these are NOO, NO, NNH, HNNO and other compounds, whose intensity of formation and decay are changed together with temperature. NO possesses the lowest strength. At 3000 K

45 YEARS OF WELDING ENGINEERING CHAIR



Figure 3. Change in molar fraction of atomic nitrogen X_i under conditions of plasma depending on molar fractions of water vapour (1-6 ---- see Figure 2)

temperature its fraction in mixture is highest, and at 5000 K it is equal almost to zero.

Dependence of molar fractions of atomic and molecular hydrogen on amount of water vapour in plasma is shown in Figure 4. At temperature up to 4000 K the molar fraction of molecular hydrogen is increased with increase in consumption of water vapour. Concentration of atomic hydrogen at 5000 K temperature is increased with increase in additions of water vapour and equal to 0.185, 0.290, 0.380, 0.475 and 0.600 at molar fractions of water vapour, respectively, 0.10, 0.17, 0.25, 0.35 and 0.50, and the molar fraction of molecular hydrogen is abruptly decreased ($2.56 \cdot 10^{-3}$



Figure 4. Change in molar fraction of atomic (*a*) and molecular (*b*) hydrogen under conditions of plasma depending on molar fractions of water vapour: $1 - H_2O = 0.10$; 2 - 0.17; 3 - 0.25; 4 - 0.35; 5 - 0.50

and $2.62 \cdot 10^{-2}$ at water adding, respectively, 0.10 and 0.50). Results of calculations showed that additions of water vapour to air reduce the partial pressure of molecular and atomic nitrogen in plasma. It should be taken into account that the presence of hydrogen in medium decreases the nitrogen content in steel [13]. The metallographic examinations of cut edges confirmed the absence of nitriding layer at water vapour additions, unlike the cutting with use of pure air plasma. In the latter case a continuous white layer of metal having microhardness up to 7000 MPa is observed at the cut edge.

Results of calculations of gas phase composition coincide almost with experimental and calculated data, given in work [8]. In addition, the method of plasma quenching was used to evaluate the results obtained [8, 14]. To model the processes of quenching, the presence of a rapid linear decrease in temperature of a plasma-chemical mixture from equilibrium to 300 K temperature for about 5 ms period and dilution of equilibrium plasma-chemical products with a water vapour in volume ratio 1:100 was assumed. The kinetic diagram remained in the previous form [10] and included 58 components in a general case.

The change was calculated as regards to a relative mass content of some plasma-chemical products in a sample, cooled down to 300 K temperature, at additions of water vapour into a plasma cutter. The relative content of plasma-chemical products represents a ratio of concentration of products by mass in a cooled sample, brought to standard atmospheric conditions, to their concentration by mass in a cooled sample of air plasma (without additions of water vapour), also brought to standard atmospheric conditions.

It was found that the yield of hydrogen-nitrides NH₃, HNO by mass in change of molar fractions of water vapour in initial mixture from 0.1736 to 0.3866 as compared with air plasma is 1.585--2.614 and 1.714--3.535 times increased, respectively. The obtained theoretical data have a good correlation with experimental data, showing that, when adding the water into a plasma jet, the concentration of ammonia and nitrides in a quenching water is 2.2 times increased, nitrites ---- 1.7 times and hydrazine ---- 2.4 times. At the same time the yield of nitrogen oxides NO and NO_2 for the same parameters is decreased, respectively, 1.23--1.62 and 1.12--1.17 times.

Owing to experimental investigations of process of pore formation in welds after air-plasma cutting of ship hull steels with addition of water to plasma, the optimum consumption of water, which depends on the air consumption, was defined. The further increase in water consumption prevents the appearance of pores with a dull colour of metal surface, caused by nitrogen, but here the pores are observed having the metal surface of a silver colour caused by hydrogen [14].

CONCLUSIONS

1. Addition of water vapour to air plasma increases the degree of conversion of a molecular nitrogen. Partial pressure of molecular and atomic nitrogen are decreased at the expense of complex compounds, whose intensity of formation depends on temperature.

2. Optimum addition of water vapour to air is an effective method of decrease in content of nitrogen in plasma, saturation of cut edges with nitrogen and increase in resistance of welds against pore formation.

3. The theoretical results obtained have a good correlation with experimental data.

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INTENSIFICATION OF THE PROCESS OF DIFFUSION BONDING OF HEAT-RESISTANT ALLOYS

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It has been established that creep processes at a transient stage used in diffusion bonding of heat-resistant alloys reduce the time of formation of a physical contact between the mating surfaces. Bonding of billets of a heat-resistant alloy in the austenised and aged states, or utilisation of spacers in a structural state differing from that of the base metal, allows intensification of the processes of interaction of the mating surfaces and provides properties of the joints at a level of those of the base metal.

Keywords: heat-resistant alloys, diffusion bonding, creep, structural state, dislocation structure, mechanical properties

Heat-resistant alloys intended for operation at high temperatures are characterised by low diffusive mobility of atoms, high creep resistance and presence of thermodynamically strong oxide films. These properties hamper a physical contact between the mating surfaces, formation of inter-atomic bonds between the surface atoms and substantially inhibit development of volumetric diffusion and recrystallisation processes needed to form a monolithic joint. This makes intensification of the diffusion bonding processes a topical problem.

The purpose of this study was to develop the flow diagram for diffusion bonding that would intensify formation of a sound bond by an example of heat-resistant alloys.

The classical flow diagram of diffusion bonding of similar metals performed in vacuum is conditionally subdivided into three stages of formation of a bond: formation of a physical contact, activation of the mating surfaces and volumetric interaction of metals joined within the bonding zone [1]. It was established that the mechanisms responsible for formation of the bonds greatly depend upon the loading and strain rates during bonding [2, 3]. Plastic strain within the bonding zone at a bonding loading rate of about 10^{-4} s⁻¹ occurs mainly by the dislocation mechanism, whereas at high strain rates the collective forms of movement of the crystalline lattice defects dominate in plastic flow [2].

Diffusion bonding of heat-resistant alloys is usually performed at low rates of plastic strain, which is realised by the steady-state creep mechanism [1]. As proved by our studies, the creep rate is 10^{-4} — 10^{-3} s⁻¹ at bonding temperatures of 1155—1195 °C and pressures of 25— 30 MPa. It is reported [4] that at the stage of the steady-state creep the equilibrium achieved between the processes of strain hardening and thermally activated weakening results in a minimum strain rate, which remains constant during the entire second stage of the creep. The process of thermal weakening is dominant at the first stage of the creep, thus ensuring higher strain rates. As the strain value grows, the rate of hardening increases and the rate of the creep decreases to its value at the second stage. In this connection, we investigated the possibility of using the first stage of the creep with higher strain rates for diffusion bonding of heat-resistant alloys. However, at the beginning we had to prove the presence of the increased strain rates, because, as reported, the onset of the steady-stage creep depends upon the temperature and composition of metal [4]. High alloys often have no stage of the steady-state creep [4, 5].

The creep rate of alloys was determined by the method described in study [6] using the IMASh-5S unit. When analysing the results, we proceeded from the assumption of a continuous and uniform mechanism of the creep process for given conditions, and determined the time dependence of the strain from the following equation [4]:

$$\varepsilon = \varepsilon_0 + \varepsilon_t (1 - e^{-rt}) + \dot{\varepsilon}_V t,$$

where ε_0 is the strain after loading; ε_t is the ultimate strain at the transitional stage; *r* is the ratio of the rate of transient creep ε_t to strain at the transient stage; *t* is the process time; and $\dot{\varepsilon}_y$ is the creep rate at the second stage.

At the same time, the rate of creep ε for any time can be described by the following equation:

$$\varepsilon = r\varepsilon_t e^{-rt} + \dot{\varepsilon}_v.$$

Processing of the experimental data for strainhardening and non-strain-hardening alloys showed a direct relationship between the creep rates at the steady-state and transient stages at the above temperatures. This relationship has the following form:

 $\varepsilon_{\rm t} = k \dot{\varepsilon}_{\rm v}$,

where *k* is the proportionality coefficient equal to 2.73--3.29 for heat-resistant alloys at temperatures of 1150--1175 °C.

The value of $\dot{\varepsilon}_y$ was found using creep constant A_c and creep activation energy ΔH_c , which were determined by the method of small temperature jumps [6].



Figure 1. Microstructure of the bond in alloy EP99 with the retained orientation of the interface relative to the initial contact plane ($\times 200$)

It was established as a result of the experiments that the strain rate in diffusion bonding of heat-resistant alloys could be substantially increased by using the transient stage of the creep. Bonding during the experiments was performed using several cycles of loading and unloading by the method suggested by E.S. Karakozov for titanium alloys [5]. As the load decreases, the strain rate also decreases, this providing the relaxation processes and leading to increase in the strain rate during the next loading cycle. Bonding was performed at temperatures of 1155--1195 °C and pressures of 25--30 MPa by varying pauses between the cycles. However, as shown by the studies, despite a substantial strain and intensive development of recrystallisation, as well as formation of a substructure. a clearly defined boundary along the former interface persists in the microstructure of the bond (Figure 1).

Electron microscopy shows that oxide films act as barriers for movement of dislocations. Thermodynamically strong and stable oxide films present on the surface of heat-resistant alloys prevent formation of intracrystalline atomic bonds within the zone of contact of the materials joined. Heating to a temperature of 1200 °C even in vacuum of 10^{-5} Pa fails to provide activation of the mating surfaces, which is attributable to dissociation and sublimation of the oxide films [7].

Examination of the surfaces of specimens in fractograms of fracture showed reduction of the physical contact formation time. Analysis of the distribution of strains within the heating zone revealed increase in the degree of deformation within the bond zone. However, formation of common grains in the bond was seen only in bonding of heat-resistant alloy EI602, which is characterised by a low content of aluminium and titanium (0.35–0.75 %). Therefore, in further studies we investigated the possibility of localising strain in the bond zone due to the shear strain development.

The conventional method of diffusion bonding with a uniform heating of both samples within the bonding zone is unfavourable in terms of surface activation [8]. As shown by the authors, in heating of cylindrical samples and formation of a barrel-like convexity with a maximal diameter in the bond zone under the effect of the compressive force, the tangential stresses and shear strains on the bond plane are equal to zero. On this basis, the authors of study [8] suggested the method for diffusion bonding of heatresistant alloys using a spacer heated to a higher temperature than that of the base metal. This leads to formation of a more favourable stress-strain state with the shear strain development. Efficiency of this method of diffusion bonding was experimentally proved. However, utilisation of a hotter spacer makes the bonding technology more complicated. Therefore, we investigated a simpler method for diffusion bonding of heat-resistant alloys, which consisted in joining alloys in different structural states, and using spacers, the structural state of which differed from that of the base metal. For example, if a strain-hardening alloy is in the austenised state, the use is made of a spacer in the aged state and vice versa. In this case, the samples assembled for bonding were simultaneously compressed and heated together with a spacer. Considering that the specific volume of metal in the aged state is smaller than in the austenised one, at the bonding (austenisation) temperatures the aged metal expands as a result of dissolution of the γ -phase. In this case, strains in the samples joined, associated with precipitation or dissolution of the γ -phase, are of a structural character. It should be noted that the volume of metal also changes in precipitation or dissolution of other redundant phases during heat treatment of strain-hardening alloys [9], while the value of structural deformations depends in many respects upon the chemical composition of heat-resistant alloys and heat treatment conditions. Structural deformations of the joined strain-hardening alloys were determined using the Chevenard dilatometer, and those of the plate samples ---- using the strain meter with a measurement base of 100 mm. Prior to the tests, the samples were subjected to water quenching from a heating temperature of 1200 °C.

To determine the temperature range in which the samples of strain-hardening alloys changed in volume under the effect of structural changes, the dilatometry patterns of heating and cooling of the samples of alloys EP539VD, EP539LM and EP99 were recorded at varied heating rates (from 5 to 20 °C/min) and holding for 1 h at temperatures of 650, 700, 750 and 850 °C. Qualitatively, the dilatometry patterns were of the same character for all the alloys studied. Quantitatively, structural deformations in a casting alloy, i.e. with a higher content of aluminium and titanium, were higher. Therefore, spacers of alloy EP539 were used for joining alloy EP99, and spacers of alloy EP99 were used for joining non-strain-hardening alloys. Deceleration of growth of length of a sample in heating took place at about 650 °C. The deformation of shortening of the samples of alloy EP99 after heating and cooling to room temperature with holding for 1 h at 850 °C was about 13.10-4.



Figure 2. Microstructure of the bond between the samples of heat-resistant alloy in different structural states: a — general view of the bonding zone of metal (EP99) in the austenised state and spacer (EP539LM) in the aged state (×300); b — fracture and displacement of interface regions together with surface oxides (dark contrast) under the effect of local shear strains (arrows) (×15,000); c — fine structure in a region of complete dissociation of the interface (×17,000); d, e — clusters of dislocations and slip systems at a distance of about 0–5 µm from the interface on the side of a spacer and base metal, respectively (×15,000); f — character of the dislocation structure in base metal at a distance of \geq 15 µm from the interface (×15,000)

Samples measuring $150 \times 60 \times 1.5$ mm were used to study the effect of temperature and time of holding of the alloys at an ageing temperature on the value of structural deformations. For this the samples were heated to 620, 650, 680, 740, 850 and 900 °C, held for 5, 15 and 30 min, 1.5, 10 and 15 h, and then examined to determine their shortening and metal hardness. As found, shortening of the samples of alloy EP99 begins as low as at 680 °C within 15--30 min. Shortening and hardness of metal were maximal at temperatures of 850--900 °C, the shortening occurring after holding for 5 min. Heating at a temperature of 620--650 °C had no effect on metal hardness, although it led to some shortening of the samples. It should be noted that also at the higher ageing temperatures the shortening effect showed up before the increase in hardness. This suggests that decrease in metal volume occurs at the first stage of ageing, where a growing diffusion of aluminium and titanium atoms begins in crystals of the oversturated solid solution. No such decrease in volume was seen in heating of the preliminarily aged samples.

The qualitatively similar results were obtained by the authors of study [9] for the hot deformation conditions and subsequent heat treatment. Changes in size of a quenched sample (especially pronounced for alloy EI828) were seen in heating at a temperature of as low as 600 °C. Also, it was established in the study that dissociation of solid solution at the ageing temperatures begins as early as after a minute holding of the quenched samples.

The determined values of structural deformations are in agreement with the X-ray diffraction analysis data on parameters of the γ -phase and oversaturated γ -solution. For example, at the γ -phase content of alloy EP539VD equal to 30 % the discrepancy between the lattice parameters is about 0.001 nm, i.e. approximately 0.3 %, while the deformation of the alloy is about 10^{-3} .

Diffusion bonding of heat-resistant alloys EP99 and EI602 in the quenched state, performed by using spacers (alloy EP539LM and EP99, respectively) in the aged state, proved intensification of the process, compared with that performed by the traditional method. Microstructure of the bond is shown in Figure 2.

Transmission electron microscopy of thin foils was conducted at different distances from the bonding line $(0-15.2 \ \mu m)$, along the bonding line and in a spacer using samples cooled immediately after pressing and after holding for 6 min, at a temperature of 1150 °C. Transmission examination of fine structure both on the side of the spacer (Figure 2, d) and on the side of the base metal (Figure 2, e) revealed a number of peculiar features of distribution of the crystalline lattice defects within the bonding zone. The central part of the spacer and base metal at a depth of more than 50 µm were characterised by a relatively low dislocation density ($\rho = (1.5-4.6) \cdot 10^9 \text{ cm}^{-2}$, Figure 2, f). Density of the slip systems grows with distance to the interface, the level of internal stresses also grows, which is evidenced by the concentration of the crystalline lattice defects in regions of interaction of the slip systems. As distance to the interface decreases, the dislocation density increases, amounting to a value of (2.0-3.5) 10^{10} cm⁻² in a zone of 5 µm deep (Figure 2, d). As the dislocation density is proportional to deformation, this proves intensification of the deformation processes due to growing shear stresses within the zone of the bond made by using a spacer,

compared with that made by the traditional method. Regions of full dissociation and migration of grain boundaries (Figure 2, c) were detected along the contact zone, resulting from development of the recrystallisation processes.

Intensive development of shear strains within the zone of the bond was definitely proved by optical examination of structure shown in Figure 2, *a*, differing from that shown in Figure 1.

The experiments confirmed efficiency of the effect of the suggested method on formation and mechanical properties of bonds in heat-resistant alloys. In bonding of alloy EP99, which was in the austenised state, through a spacer of alloy EP539LM in the aged state, the specimens fractured at a distance from the bond both in short- and long-term tests at a temperature of 900 °C. Bonding of samples of the alloys was performed by heating in the compressed state under a pressure of 15 MPa.

Similar results were obtained in bonding of alloys EP99 and EI602. In this case the compressive force at a bonding temperature was decreased by a factor of 1.65, compared with bonding using no spacers (15 and 25 MPa, respectively). Bonding of alloy EP99 in the as-received state performed at a pressure of 15 MPa provided bonds with a tensile strength of 66--133/92 MPa at a temperature of 900 °C. Bonding of the samples, one of which was in the austenised state and the other in the aged state, under the same conditions provided bonds with a tensile strength of 550--630/583 MPa. In bonding of the austenised samples of alloy EP99 through a spacer of the same alloy the tensile strength of the bonds was 540--652/592 MPa. The bonded samples of alloy EI602 with a spacer of alloy EP99 in the aged state always fractured in alloy EI602, whereas the samples bonded under the same conditions but without a spacer always fractured along the bonding line. Therefore, the investigations of structure and properties of the bonds prove the substantial effect of the bonding method considered on the intensity of formation and quality of the bonds.

CONCLUSIONS

1. Using the transient stage of creep of heat-resistant alloys allows the rate of plastic strain to be increased

and time of formation of a physical contact between the mating surfaces in diffusion bonding to be reduced. However, in this case the microstructure of the bonds in strain-hardening alloys retains a clearly defined grain boundary along the former bonding line.

2. Activation of mating surfaces, localisation of strains within the bond zone and intensive development of volumetric processes in diffusion bonding of heat-resistant alloys can be achieved by using of spacers that are in a structural state differing from that of the base metal.

3. Dissolution of the hardening phase in heat-resistant alloys and a corresponding increase in the specific volume of the spacer material lead to increase in level of tangential stresses and development of shear strain within the zone of contact between the samples joined, thus allowing intensification of the processes of interaction of the mating surfaces and providing properties of the bonds at a level of those of the base metal.

4. The use of billets of heat-resistant alloys in different structural states allows specific compressive pressures in diffusion bonding to be decreased and provides a high quality of the bonds.

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ALLOWANCE FOR PLASTIC STRAINS IN COMPUTER MODELING OF STRESS FIELDS IN BRAZED JOINTS MADE FROM DISSIMILAR MATERIALS

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Using the finite element method, the procedure of modeling fields of natural stresses and strains with allowance for plastic strains has been developed. A good accuracy of the method is shown by comparison of theoretical and experimental values. It was found that the first plastic strains are appeared in the joint at its edges where high tangential τ_{xy} and normal σ_{v} stresses are combined.

Keywords: brazing, dissimilar materials, plastic strains, stresses, modeling, finite element method

Application of dissimilar materials in assemblies of machine-building and power engineering equipment provides an optimum combination and distribution of service properties between separate parts of the structure. Combinations of metals with ceramics, refractory, composite and other materials have found here the widest spreading.

Brazing is considered to be the most effective method of joining the dissimilar materials, however, the assurance of joint performance at different physical-chemical and physical-mechanical properties of materials and alloys being joined remains still the main problem. Therefore, the investigation of stressstrain state of assemblies made from dissimilar materials is actual.

In spite of numerous investigations, devoted to dissimilar material joining, some research and applied problems remain until now little-studied and require their solution. In particular, a very small amount of systematized data is available in technical literature about the effect of physical-mechanical properties of materials, geometric characteristics of structures on formation and distribution of natural stresses and strains in brazed joints. The approach was developed earlier which allowed modeling the fields of natural stresses and strains in structures made from dissimilar materials without allowance for plastic strains [1]. In work [2] the natural deformations of specimen of a three-layer bar type were investigated experimentally. Comparison of experimental and calculated data showed a good accuracy of the method developed. If one of the materials is ductile, then the plastic strains, initiated in it, influence the pattern of the stress-strain state.

The aim of the present work is to develop the method of computer modeling of fields of stresses and strains in assemblies made from dissimilar materials with allowance for plastic strains, and also to study the effect of plastic strains on the field of stresses in a two-layer bar. In general case of elastic-plastic deforming, the deformations in any point can be presented in the form [3]

$$\varepsilon = \varepsilon_e + \varepsilon_p + \varepsilon_d$$

where ε_e is the elastic strain; ε_p are the strains of a spontaneous plasticity; ε_d are the strains of a diffusion plasticity (creep).

Strains of a spontaneous plasticity were calculated by the theory of elastic-plastic strains, the material was assumed to be perfectly ductile [3]. Conditional diagram of tension is shown in Figure 1. Strains of a diffusion plasticity were not taken into account.

The appearance of plastic strains in each finite element (FE) was determined from Mises--Henke condition [3]:

$$\sigma_i > \sigma_{\rm YS}$$
,

where $\sigma_i = \frac{1}{\sqrt{2}} \sqrt{(\sigma_x - \sigma_y)^2 + (\sigma_x - \sigma_z)^2 + (\sigma_y - \sigma_z)^2 + 6(\tau_{xy}^2 + \tau_{xz}^2 + \tau_{yz}^2)}$ is the intensity of stresses, in case of plane stressed state

 $\sigma_i = \sqrt{\sigma_x^2 + \sigma_y^2 - \sigma_x \sigma_y + 3\tau_{xy}^2}.$ In fulfillment of condition $\sigma > \sigma_{\text{YS}}$, calculation in some FE was repeated with a secant modulus $E_p =$

some FE was repeated with a secant modulus $E_p = E(1 - p)$, where p is the parameter characterizing the degree of propagation of plastic strains in appropriate FE. Calculation was made using the method



Figure 1. Scheme of determination of plastic strains: 1 ---- calculation in the first approximation at initial modulus of elasticity E; 2 — calculation with allowance for plastic strains with a secant modulus E_n

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of successive approximations with a constant increase in *p*. Initial value *p* = 0 was increased successively at 0.01 pitch until fulfillment of condition $\sigma \leq \sigma_{\rm YS}$, i.e. the absence of yielding in all FE. The intensity of plastic strains in plastically deformed FE was calculated by the same approximate method used for linear stressed state.

As follows from Figure 1, the total strains $\varepsilon = \varepsilon_p + \varepsilon_e = \sigma_{YS} / E_p$. As $\varepsilon_e = \sigma_{YS} / E_p$, $E_p = E(1 - p)$, then after substituting and transformations we shall obtain

$$\varepsilon_{\rm p} = \frac{\sigma_{\rm YS}}{E} \frac{p}{1-p}$$

Elastic strains and stresses were calculated with allowance for final values of a secant modulus in FE after plastic deforming.

To check the accuracy of calculation with allowance for plastic strains we calculated sagging of the specimen, described and studied experimentally in work [4].

Analysis of results of modeling showed a good correlation of values (from results of modeling f = 2.1 mm, according to work [4] f = 2.5 mm).

Using the method developed, the effect of plastic strains on fields of natural stresses was investigated for several variants of joints of a symmetrical twolayer bar ($B_1 = B_2$). Assemblies were studied with similar moduluses of elasticity and yield strength of layers ($E_1 = E_2$, $\sigma_{YS1} = \sigma_{YS2}$), similar moduluses ($E_1 = E_2$), but different values of yield strength of layers $(\sigma_{YS1} > \sigma_{YS2})$, moduluses of elasticity $(E_1 > E_2)$ and yield strengths of layers $(\sigma_{YS1} > \sigma_{YS2})$. Coefficients of temperature expansion of materials joined were taken different by 2 times $(\alpha_1 = 2\alpha_2)$.

Degree of propagation of plastic strains varied by changing σ_{YS} from 375 to 100 MPa so, that plastic strains were propagating simultaneously in both layers. Results of solution were compared with those of «elastic» solution ($\sigma_{YS} = 400$ MPa).

To have a quantitative evaluation of plastic strains on the level of natural stresses, the conception «relative elasticity of material» ($e = \sigma_{YS} / \sigma_{max}$ is the ratio of yield strength of material to maximum stresses in it, determined by «elastic» solution) and coefficient of effect of plastic strains ($k_p = \sigma_{max}^p / \sigma_{max}$ is the ratio of maximum stresses with allowance for plastic strains to stresses at elastic solution) were used.

Figure 2 shows the fields of equivalent stresses σ_i in upper layer on the right from vertical axis of symmetry *y* of the symmetrical two-layer bar ($E_1 = E_2$, $\sigma_{YS1} = \sigma_{YS2}$) at different values of relative elasticity of material $e = \sigma_{YS} / \sigma_{max}$. As is seen from Figure 2, the first plastic strains (regions *A*) are appeared at relative elasticity e = 1.25, i.e. $\sigma_{max} < \sigma_{YS}$, in the zone of joint at a small distance from edge (x = 0.8-0.9L; *L* is the length), where high longitudinal σ_x and tangential τ_{xy} stresses are combined. With decrease in relative elasticity of the material, the zone of plastic strains is increased, thus propagating gradu-



Figure 2. Fields of intensity of stresses σ_i at $\sigma_{\rm YS} = 375$ MPa, e = 1.25 (*a*); $\sigma_{\rm YS} = 350$ MPa, e = 1.17 (*b*); $\sigma_{\rm YS} = 300$ MPa, e = 1 (*c*); $\sigma_{\rm YS} = 100$ MPa, e = 0.3 (*d*). Regions A correspond to $\sigma_i \ge \sigma_{\rm YS}$



ally along the joint and transferring to the free surface (e < 1).

Diagrams of stresses for case $E_1 = E_2$, $\sigma_{YS1} = \sigma_{YS2}$, are presented in Figure 3. Plastic strains reduce the level of stresses σ_x only in most stressed points near the joint (Figure 3, *a*), while in far points the stresses are little changed. In length of the joint the degree of reduction of maximum stresses in the zone of plastic strains is distributed non-uniformly (Figure 3, *b*).

At the first moment of appearance of plastic strains, the stresses σ_x are decreased only near the free lateral surface (at the joint edges), i.e. in the zone with high tangential stresses. In the middle part of the joint, the maximum stresses σ_x are not changed in this case. At σ_{YS} decrease the zone and degree of plastic strains are increased, maximum σ_x are reduced in the entire length of the joint, but they are decreased more intensively at the joint edges (Figure 3, *b*).

Dependence of maximum stresses σ_x on relative elasticity of material in case $E_1 = E_2$, $\sigma_{YS1} = \sigma_{YS2}$ is shown in Figure 4, *a*. Noticeable reduction (by 25–30 %) of σ_x at the joint edge was recorded even at e = 1; at e = 0.5 the reduction reaches 2 and more times ($k_p = 0.5$ in the middle and 0.25 at the joint edge).

Taking into account that the zone of increased plastic strains occupies a comparatively small part of

the joint it is possible to consider approximately e = 1 in the region $k_p = 1$ with an error into a safe zone at practical calculations, i.e. the effect of plastic strains can be not taken into account and at e < 1 the relationship can be taken linear $k_p = e$.

Diagrams of transverse stresses σ_y are shown in Figure 3, *c* and *d*. As is seen from diagrams, the stresses begin to decrease just after appearance of the first plastic strains and they are decreased in all points both on external surface (Figure 3, *c*) and also along the interface (Figure 3, *d*).

Dependence of degree of reduction of maximum stresses on relative elasticity of materials is shown in Figure 4, b. Significant reduction in maximum stresses σ_y is started at relative elasticity of materials e = 0.8. At e = 0.5 the coefficient of effect of plastic strains on maximum stresses σ_y is $k_p = 0.5$.

Evidently, to make an approximate assessment of effect of plastic strains on stress σ_y it is possible to consider $k_p = 1$ in practical calculations at $e \ge 0.9$ and $k_p = 1.1e$ at $e \le 0.9$ (dashed line in Figure 4, *b*).

Diagrams of tangential stresses τ_{xy} are shown in Figure 3, *e* and *f*. Here, the tangential stresses are decreased greatly in all points, but only if maximum normal stresses σ_x reach σ_{YS} , i.e. when the zone of plastic strains is spread to the larger part of the joint.

Dependence of degree of reduction of maximum tangential stresses τ_{xy} in the joint on relative elasticity



Figure 3. Diagrams of stresses σ_x (*a*, *b*), σ_y (*c*, *d*) and τ_{xy} (*e*, *f*) across (*a*, *c*, *e*) and along (*b*, *d*, *f*) the joint at interface (*y* = 0): 1 --- σ_{YS} = 400; 2 --- 350; 3 - 200; 4 - 100 MPa



Figure 4. Dependence of maximum stresses σ_x (a), σ_y (b) and τ_{xy} (c) in middle section (x/L = 0) (1) and at the joint edge (x/L = 0.9) (2) on relative elasticity of material

of the material (Figure 4, *c*) is very close to that for stresses σ_y . Therefore $k_p = 1$ can be also considered for τ_{xy} at $e \ge 0.9$ and $k_p = 1.1e$ at $e \le 0.9$ (dashed line).

Analysis of results of modeling for cases with similar modules ($E_1 = E_2$), but different values of yield strength ($\sigma_{YS1} > \sigma_{YS2}$), and also different values of modulus of elasticity ($E_1 > E_2$) and yield strength ($\sigma_{YS1} > \sigma_{YS2}$) of layers showed that, as a whole, the distribution of stresses is governed by the regularities obtained for the case mentioned. From dependencies of maximum stresses on relative elasticity of material the following regularities for assessment of effect of plastic strains were obtained.

For case with similar moduluses $(E_1 = E_2)$, but different yield strengths of layers $(\sigma_{YS1} > \sigma_{YS2})$:

• effect on stress σ_x in a layer with high value σ_{YS} (without plastic strains) $k_p = 1$ at $e_2 \ge 0.5$ and $k_p = 2e_2$ at $e_2 \le 0.5$, in a layer with lower σ_{YS} (with plastic strains) $k_p = 1$ at $e_2 \ge 1$ and $k_p = e$ at $e_2 \le 1$;

• effect on stress σ_y with allowance for plastic strains of the second layer can be taken in elastic layer $k_p = 1$ at $e_2 \ge 0.55$ and $k_p = 1.8e_2$ at $e_2 \le 0.5$, and in plastic layer $k_p = 1$ at $e_2 \ge 0.9$ and $k_p = 1.1$ at $e_2 \le 0.9$;

• effect on stress τ_{xy} , $k_p = 1$ at $e_2 \ge 0.9$ and $k_p = 1.1e_2$ at $e_2 \le 0.9$.

For case with different moduluses $(E_1 > E_2)$ and different yield strengths of layers $(\sigma_{YS1} > \sigma_{YS2})$:

• effect on stress σ_x for material with high value E and yield strength in the region $e \ge 0.4 k_p = 1$, i.e. the effect plastic strains can be neglected, and at $e < 0.4 k_p = 2.5e$ should be taken;

• for material with a low value *E* and yield strength at $e \ge 0.7 k_p = 1$, and at e < 0.7 should be taken $k_p = = 1.43e$;

• effect on stress σ_y in elastic material $k_p = 1$ at $e \ge 0.65$ and $k_p = 1.54e_2$ at e < 0.65, in plastic material $k_p = 1$ at $e \ge 0.4$ and $k_p = 2.5e_2$ at e < 0.4;

• effect on stress τ_{xy} , $k_p = 1$ at $e \ge 0.65$ and $k_p = 1.54e$ at e < 0.65.

CONCLUSIONS

1. Method of computer modeling of fields of natural stresses and strains in assemblies from dissimilar materials with allowance for plastic strains has been developed. The accuracy of method was confirmed by the comparison of results obtained with experimental data.

2. Effect of plastic strains on the distribution and level of maximum stresses in two-layer symmetrical bar at different combination of properties of materials was analyzed. It was found that the first plastic strains are appeared at the joint edge, where high values of tangential τ_{xy} and normal σ_x stresses are combined, here, σ_x^{max} do not reach the yield strength. It is shown that negligible plastic strains have a low effect of stress fields as a whole. The noticeable reduction in maximum stresses was observed at relative elasticity *e* of lower than 0.9--1.0.

3. Simple relationships are offered for approximate evaluation of plastic strains.

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APPLICATION OF A NONCONSUMABLE CARBON-ELECTRODE ARC AT SURFACE TREATMENT OF STEELS WITH A LOW HARDENABILITY

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Structural transformations in low-carbon steels during carbon-electrode arc surface hardening by heating without melting are considered. The effect of carbon, grain size and initial metal temperature on the conditions of hardening of low-carbon steels is shown.

Keywords: carbon electrode, surface treatment, low-carbon steels, hardenability, martensite

Improvement of performance of parts of machines and structures operated under the conditions of abrasive wear, is achieved by applying on their surfaces special wear-resistant coatings or quench hardening of the surface layers of steel.

Technologies using concentrated local heat sources have become widely accepted lately for surface hardening of steel [1--7]. The latter are based on formation of an austenitic structure during surface heating and its transformation into martensite or bainite at subsequent cooling with formation of a structure characterized by a pronounced refinement of grains, laths, packs of martensite plates and presence of disperse carbides.

One of the simplest methods to implement the surface hardening is use of a nonconsumable-electrode arc as a heat source. Such modes of treatment include plasma treatment, as well as treatment by a tungstenand carbon-electrode arc.

Most of the work on surface hardening has been done using plasma heating with melting of the surface being hardened. Analysis of hardening processes arising in carbon steels as a result of heating without surface melting is given little attention in publications. Treatment without surface melting is preferable from the viewpoint of technology, but the hardened zone structure has its special features, due to the short time of the upper layers being at temperatures above A_{cs} .

The most readily available local heat source is an arc discharge from a carbon or graphite electrode.

Table 1. Modes of heating cylindrical steel samples by a carbonelectrode arc

Steel grade	I _w , A	$U_{ m a}$, V	v _h , m∕h	$q_{ m ef}$, W	q _h , J∕cm
40Kh	85	20	225	1020	162
18KhGT	75	23	225	1035	165
30KhGSA	75	30	225	1350	215

Such a heat source is cost-effective for surface hardening of steels, as the straight polarity arc runs in air atmosphere in a stable manner even at a low discharge power, effective efficiency of heating is about 60 %, and the electrodes have a high enough wear resistance.

Influence of heating by an arc from 8 mm carbon electrode without surface melting on structural transformations in low-alloyed steels 40Kh, 18KhGT and 30KhGSA was studied on cylindrical samples of 20 mm diameter with preheating along a helix. Rotation frequency was 60 rpm and speed of parallel displacement of the arc was 1.25 mm/s. Preheating modes are given in Table 1. After preheating the surface layers develop a martensite or martensite-bainite structure with a certain carbon heterogeneity. This is due to insufficient time of the metal staying in the region of the austenitic condition, at which carbon could diffuse into the centers of the former ferrite grains. This is particularly noticeable for 18KhGT steel, in which the average size of the ferrite grain was the largest before heating.

Investigations of hardness distribution in hardened surface layers (Figure 1) showed that the depth of the hardened layer is down to 0.5--0.8 mm.

The highest hardness of the surface layer (about HV 500) is observed in steel 40Kh. In 30KhGSA steel the layer with HV 450 hardness has the depth of 0.8 mm. The surface layer of 18KhGT steel has the hardness of more than HV 400. Thus, performed research showed the possibility of using a carbon-elec-





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Figure 2. Mictrostructure of steel 10: a — initial condition; b — after heating by a carbon-electrode arc (×330)

trode arc for heat treatment of structural steels. It is characterized by a low power consumption and is readily controllable.

For unalloyed and low-alloyed low-carbon steels a significant increase of the hardness of surface layers can be achieved ensuring a comparatively uniform distribution of carbon in the austenite formed at heating with subsequent very fast cooling. Investigation of heating by a carbon-electrode arc of the surface of low-carbon steel plates with an initial ferrite-pearlite structure demonstrated that melting of the steel surface starts at heat input of more than 500 J/cm. Steel 10 with a very small amount of pearlite was used to study the diffusion processes and structural transformations at heating with such a heat input. Microstructure of base metal of this steel is given in Figure 2, a. After arc treatment, the heated metal develops a more complex structure. In the near-surface layers heated up to maximum temperatures of 1200--1300 °C, it consists of comparatively coarser ferrite grains, which have been completely recrystallized, and dark areas, made up predominantly by bainite (Figure 2, b). No precipitation of free hypoeutectoid ferrite grains is observed inside the dark areas of the structure. Values of microhardness determined in a microhardness meter PMT-3 at 50 g load in one of the dark areas of the structure vary from the center to the periphery (HV 360, HV 215, HV 175, respectively). Microhardness of ferrite grains is HV 140. In the center of individual dark areas the maximum microhardness value is HV 550. The above regularities

of hardness variation in the dark areas of the structure are due to lower concentration of carbon in the regions at greater distance from the center of the former pearlite grains. It is characteristic of the bainite structure, formed from austenite with different concentration of carbon. Hardness value of HV 550 corresponds to that of martensite, formed from austenite with the carbon content of 0.35--0.40 %. For heat input of 500 J/ cm at heating the average path of carbon diffusion for high-temperature regions of the HAZ metal is about 0.015--0.020 mm.

Overlapping of the boundaries of carbon diffusion from initial pearlite grains into the ferrite grains will depend on the ratio of the volumes of ferrite and pearlite in the initial structure, as well as average values of pearlite grains size. Formation of a continuous frame of the martensite structure can be ensured only at the carbon diffusion path longer than the distance from the pearlite grain boundary to the cube corner. In this case a significant increase in the hardness of steel is achieved.

At evaluation of the minimum value of path length of carbon diffusion for steel with an initial ferritepearlite structure we assume that the pearlite grains have a spherical shape and are uniformly distributed across the metal volume. Let us divide the entire volume into elementary cubes with edge *A*, where each cube takes up one pearlite grain, located in the cube center. Value of the side of such a cube will be found from the following ratio:

$$A = D \sqrt[3]{\frac{\pi}{6\beta}},$$
 (1)

where *D* is the average diameter of the spherical pearlite grain; β is the volume fraction of pearlite in steel.

Such a model of the texture of low-carbon steel with an initial ferrite-pearlite structure allows evaluation of the minimum path length of carbon diffusion in austenite L_d , depending on the volume fraction of pearlite and mean diameter of pearlite grain. In the assumed model the distance from the pearlite grain boundary to the cube corner is determined for the specified ferrite-pearlite structure. It can be found from the following expression:

$$L_{\rm d} = \frac{D}{2} \left(1.73 \sqrt[3]{\frac{\pi}{6\beta}} - 1 \right).$$
 (2)

Figure 3 gives the calculated data of the dependence of minimum diffusion path of carbon atoms with the initial ferrite-pearlite structure of low-carbon steels with carbon content of 0.1 and 0.2 % on the mean grain size of pearlite. Calculation of volume fraction of pearlite β was conducted allowing for part of carbon being in the ferrite. Value of maximum carbon solubility in ferrite of 0.02 wt.% was assumed. This also shows the position of the grain point in the assumption that the mean size of pearlite grain corresponds to its average value.

OURNAL

It is experimentally established that any significant surface hardening of low-carbon steel with an initial ferrite-pearlite structure due to its heating by a carbon-electrode arc without melting can be achieved only for fine-grained steels. At carbon content of about 0.2 % treatment of steel with grain point 7 and higher is effective. With lowering of carbon content in steel the critical length of the diffusion path becomes longer, and the steel surface hardening can be achieved only for super fine-grained steels. The above value of the carbon diffusion path of 0.015--0.020 mm corresponds to the critical value of heat input of 500 J/cm, above which the surface melts. Considering the possibility of variation of the arcing modes, as well as the need for the heating zones to overlap to achieve a uniform depth of the hardened layer, which is produced on the surface of the already heated metal, the heat input of heating should be below the critical energy, this leading to a shorter path of carbon diffusion in austenite.

Thus, the initial ferrite and pearlite structure should be fine-grained in low-carbon steels to achieve a significant hardening of its surface at its one-time heating by a carbon-electrode arc. A mandatory condition for producing a high hardness in the surface layer of low-carbon steels with an initial ferrite-pearlite structure is achieving such a condition at heating, when carbon, concentrated in pearlite grains, has enough time for spreading uniformly across the entire volume of austenite. If the time of austenite staying at the temperature above A_{c_s} at such a one-time heating is insufficient, then in order to achieve an approximately uniform redistribution of carbon in austenite, it is necessary either to select steel with a super fine initial structure of ferrite and pearlite, or perform additional preparation of the surface layers of coarsegrained steel, providing a more uniform distribution of carbon-containing phases across the volume. Such a distribution of carbon in the steel volume can be achieved by additional heating without melting of the low-carbon steel surface with subsequent fast cooling. In this case the cooling mode should suppress the pearlite transformation of austenite. The number of such heating operations for this steel can be determined experimentally, when at final martempering a hardness value in the surface layer is achieved, which is close to martensite hardness at a average content of carbon in this steel.

Required cooling modes for transformation of the HAZ metal austenite into martensite can be evaluated by calculation procedures. For HAZ metal regions heated to maximum temperature of 1200--1300 °C (on condition of carbon distribution in austenite close to a uniform one), a critical cooling mode, at which austenite transforms into a structure consisting of 95 % martensite and 5 % bainite, can be calculated by the following formula [8]:



Figure 3. Influence of the mean pearlite grain size on minimum diffusion path of carbon in austenite of low-carbon steels with an initial ferrite-pearlite structure: 1 -steel with relative fraction of pearlite $\beta = 0.2$ (0.1 % C); 2 — $\beta = 0.225$ (0.2 % C)

where t_3 is the time of austenite cooling from critical point A_{c_i} to the temperature of the start of martensite transformation $M_{\rm s}$; C is the carbon weight percentage in austenite; r = 0.32(Si - 0.3) + 0.62(Mn - 0.6) ++ 0.66(Cr - 0.15) + 0.25(Ni - 0.15) + 0.1Mo + 1.08V.

Time Δt_3 , calculated by formula (3) is equal to 1 s for steel St3. Full hardening of austenite in steel can proceed, if the time of cooling from A_{c_s} to M_s at the stage of cooing after arc heating $\Delta t_{\rm c}$ is less than Δt_3 .

Time $\Delta t_{\rm c}$ depends on the heat input of heating and values of critical points A_{c_3} and M_s and is evaluated by the following formula:

$$\Delta t_{\rm c} = \frac{q_{\rm h}}{2\pi\lambda} \left(\frac{1}{M_{\rm e} - T_0} - \frac{1}{A_{c_3} - T_0} \right)$$
(4)

where q_h is the heat input of heating; λ is the coefficient of steel heat conductivity; T_0 is the initial temperature of the heated surface of the steel body.

Critical points A_{c_3} and M_s are found from the following formulas [9]:

$$A_{c_{3}} (^{\circ}\tilde{N}) = 910 - 229\tilde{N} + 32Si - 25Mn - 8Cr - 18Ni + 2Mo + 117V - 24Cu + 7W - 120B; (5)$$

$$M_{s} (^{\circ}\tilde{N}) = 550 - 400C - 6Si - 33Mn - 25Cr - 16Ni + 27Mo + 130V - 17Cu - 15W - 200B.$$

Calculations performed for the case of heating without melting by a nonconsumable-electrode arc discharge at heat inputs of 500 and 250 J/ cm of steels 10 (0.1C), 15 (0.15C), 20 (0.2C) and 09G2 (0.09C) with element content by an arithmetical mean composition, envisaged by the standards (wt.%), demonstrated the actual possibility of their surface hardening (Table 2).

In all the considered cases the time of cooling from critical point A_{c_3} to temperature M_s of austenite of the surface layer is less than Δt_3 in low-carbon steels heated up to 1200--1250 °C by the heat of a carbonelectrode arc, in which an approximately uniform distribution of carbon is achieved. Since during continuous treatment the heat applied by an arc heats the item, it is necessary to determine the influence of this self-heating on formation of the sought structure at intermediate heating, required for uniform redistri-

$$\Delta t_3 (\mathbf{c}) = (5.5 \text{ C})^{1.45} \cdot \mathbf{10}^r, \qquad (3)$$

(2)

GOURNAL

Table 2. Calculated data of critical points and hardenability parameters of low-carbon steels at their heating by an arc discharge without melting

Steel grade	A_{c_3} , °C	\check{I} , ${}^{\hat{1}} ilde{N}$	$\Delta t_{\rm c}$, s	Δt_3 , s	$rac{\Delta t_{50ec{1}}}{s}$,	$T_{ m 50 \check{l}}$, ${}^{\hat{1}} ilde{N}$
10	876	483	0.20/0.10	0.35	0.68	287/369
15	865	463	0.21/0.11	0.61	1.19	336/392
20	853	443	0.23/0.12	0.92	1.78	352/394
09G2	850	450	0.22/0.11	1.47	3.83	404/426
Note. 500 J∕c	The nu m, and th	umerator he denom	gives the inator — up	values to 250 J	for hea J∕cm.	t input of

bution of carbon and final complete martempering. T_{50M} can be taken as the maximum possible temperature at intermediate heating, which yields a structure with 50 % martensite. Values of these temperatures for the considered steels and heat inputs of heating are calculated from the expressions given in [8]. In addition, Table 2 gives the design time of cooling from A_{c_3} to M_s , when a structure with 50 % martensite forms from the austenite. This data indicates that the heating operations designed to provide a uniform distribution of carbon, can be performed continuously without forced cooling of the item to temperatures below T_{50M} .

Influence of the part temperature on the final hardening process requires further analysis. For this purpose the expressions given in [8] and (4) were used to calculate the influence of the initial steel temperature on the amount of martensite at heating at the heat input of 500 and 250 J/cm. It is found that heating of unalloyed low-carbon steels 10, 15 and 20 up to 200, 250 and 300 °C at heat input of 250 J/ cm leads to formation of a fully martensite structure in the steel layers heated up to high temperatures. Only for steel 10 at metal temperature of 300 °C the amount of martensite is 85 %. Influence of carbon content and initial temperature on the amount of formed martensite in the heated layers of these steels at the heat input of 500 J/cm is shown in Figure 4. From this Figure it follows that in steel 20 the fully martensite structure forms even at its preheating up to 300 °C. For steels with a lower carbon content heating up to



Figure 4. Influence of carbon content in low-carbon unalloyed steels on the amount of martensite in a hardened surface layer after heating by an arc discharge with the heat input $q_h = 500 \text{ J}/\text{ cm}$ at different initial temperatures of the metal: $1 - T_0 = 20$; 2 - 200; 3 - 250; 4 - 300 °C

250--300 °C leads to incomplete hardening of the heated layers. As the heat input of 500 J/ cm at surface hardening without melting is the maximum possible one, the performed analysis shows that the considered hardening method can be combined with low-temperature tempering. In this case, part heating up to the required tempering temperature is achieved due to the arc heat, designed to produce the surface layer structure and its subsequent hardening. Such a method of surface hardening is characterized by heating being followed by abrupt cooling to the temperature of the end of martensite transformation, subsequent soaking at tempering temperatures and slower cooling of the part after the end of its treatment. This promotes relaxation of inner structural stresses in martensite with preservation of the high values of hardness. As regards surface hardening of low-carbon low-alloyed steel 09G2, 100 % of martensite forms in it in all the considered cases after complete heating.

Low-carbon and low-carbon low-alloyed steels have a high temperature of the start and end of martensite transformation and in combination with low-temperature tempering have a high resistance to initiation of hardening cracks, despite the cooling rates of about 10^{-3} °C/s. Temperature of the end of martensite transformation is above that of low-temperature tempering, which does not require their complete cooling right after hardening. So, for the considered steels the temperature of formation of 75 % martensite, calculated by the equation given in [10], is 450 for steel 10, 420 for steel 15, 390 for steel 20 and 410 °C for steel 09G2.

Analysis data are readily confirmed by microstructural studies of the hardened surface layers of steel St3 with 0.16 % carbon, made in the scanning electron microscope-microanalyzer REMMA-102 (Figure 5).

The above statements were checked by repeated heating of the surface of steel St3 by a carbon-electrode arc at heat input of about 500 J/cm. One-time heating of the surface of this steel (Figure 5, b) leads to increase of hardness at the depth of 0.2 mm to HV 270, which is by HV 120 higher than that of the base metal.

Two-times heating increases the diffusion path of carbon in steel due to a longer time of austenite staying in the temperature region above critical point A_{c_3} , and after cooling leads to higher hardness of the surface layer at the depth of 0.2 mm already of up to HV 380. However, as is seen from Figure 5, c, in addition to martensite and bainite, also a certain amount of ferrite grains is observed in the surface layer after such heating. This is indicative of a still incomplete equalizing of carbon concentration in austenite in the entire volume of the heated layer. Three-times heating leads to formation of a predominantly martensite structure in the surface layer of steel St3. However, even after such a treatment individual islets of ferrite are found in the microstructure of the hardened layer (Figure 5, d). Three-times heating ensures increase of hardness of steel St3 up to HV 400 at the depth of 0.3 mm,



Figure 5. Microstructure of St3 (0.16 % C): a — base metal; b — surface layer after one-time heating by a carbon-electrode arc at $q_h = 500 \text{ J/cm}$ (HV 270); c — after two-time heating (HV 380); d — after three-times heating (HV 400) (×1500)

and up to HV 360 at the depth of 0.6 mm. Such a hardness of the surface layer is practically identical to the maximum possible value for martensite with carbon content close to its average content in the studied steel. In this case a more abrupt transition through the structure of the hardened layer to the base metal is also observed, compared to such a transition at one-time heating.

Comparison of microstructures shows that an increase of the acicular component of martensite is observed with the increase of the number of heating operations. After three-times heating a practically complete martempering of steel is achieved in the structure of the surface layer at the depth of 0.3 mm.

The process of surface hardening of low-carbon steels with an initial ferrite-pearlite structure can be applied both to flat surfaces, and to surfaces of cylindrical parts. In this case a hardened strip of about 0.7 mm depth and 4 mm width is obtained in a plane after one-time heating. In order to treat the entire plane, the next passes are performed with arc shifting by 2.5--3.0 mm. Heating of the cylindrical part is performed at its rotation by a helix with the indicated pitch. The required heating rate in the specified arcing mode is selected from the condition of the heat input being less than 500 J/ cm to avoid steel melting. During heating it is further necessary to note that the treated metal temperature rises due to self-heating. Therefore, during treatment it is necessary to avoid the part temperature rising above 300 °C. If the part dimensions do not permit satisfying this condition,

further heating should be conducted with application of artificial cooling or with pauses for natural cooling of the metal.

Developed method [11] of surface hardening of steels by their heating without melting by a nonconsumable carbon-electrode arc widens the range of hardenable steels without changing the technology of their heating. The difference consists in that for medium-carbon steels quench hardening of the surface layers is achieved at one-time heating, and for lowcarbon steels with an initial ferrite-pearlite structure ---- by multiple heating. In the latter case the number of heating operations is determined by the specific content of carbon in the low-carbon steel and its grain point.

Hardening surface treatment by a nonconsumableelectrode arc heating can be applied not only in fabrication of parts and structures, but also after reconditioning of worn parts by surfacing. Surfacing can be performed by low-carbon and low-carbon low-alloyed wires, providing a high resistance of the deposited metal to cold and hot cracking. The ferrite-pearlite structure of the deposited metal, obtained when using such surfacing wires can be hardened in the surface layer by additional heating of the machined part surface by a nonconsumable-electrode arc.

CONCLUSIONS

1. It is found that for heating without melting of the steel surface layers by a carbon-electrode arc the heat input should not exceed 500 J/cm.

2. A significant increase of hardness of low-carbon steels with the initial ferrite-pearlite structure after one-time heating without melting by a nonconsumable-electrode arc can be achieved only for finegrained steels with grain point 8 and higher.

3. For coarse-grained low-carbon steels achievement of a high hardness in surface hardening by heating by a nonconsumable-electrode arc requires preliminary conditioning of the surface layer.

4. High hardness of the surface layers at hardening is due to additional heating of the surface to achieve a more uniform redistribution of carbon through the entire volume of the metal and to produce martensite after final heating with the average carbon content close to its content in this steel.

5. Low-carbon steels have a high temperature of the end of martensite transformation, therefore their surface hardening by heating by a nonconsumableelectrode arc can be conducted without forced cooling. The heat consumed in item heating can be simultaneously used for low-temperature tempering. For unalloyed low-carbon steels the maximum temperature of the hardened item should not be higher than 300 °C.

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EFFECT OF RESIDUAL STRESSES ON TECHNOLOGICAL STRENGTH OF WELDED JOINTS OF HIGH-STRENGTH STEEL 14KhGN2MDAFB

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Effect of diffusible hydrogen content and welding heat input on delayed fracture resistance of 14KhGN2MDAFB steel welded joints made by using wire Sv-10KhN2GSMFTYu, has been studied. Relationship has been established between the level of residual stresses and longitudinal and transverse cold crack resistance of the joints in the above steel welded using wires that provide the weld metal with bainitic-martensitic (BM), ferritic-bainitic (FB) and austenitic (A) structures.

Keywords: arc welding, high-strength low-alloy steels, welded joint, cold cracks, residual welding stresses, diffusive hydrogen, heat input of welding, weld metal, heat-affected zone

High-strength steels of 600--1000 MPa and higher yield strength find a wide spreading in manufacture of welded structures. The use of these steels makes it possible to decrease significantly the mass of structures, to increase their service characteristics, and also to widen the technical capabilities in creation of mechanisms of machines and engineering constructions. One of the most complicated problems in welding of the above-mentioned steels is associated with prevention of a delayed fracture in welded joints.

Initiation of cold cracks in high-strength steel welded joints is defined by such factors as content of diffusive hydrogen $[H]_{diff}$, the presence of structures, reducing their resistance to brittle fractures, and residual stresses [1--4].

The aim of the present work was to widen the knowledge in effect of residual welding stresses on the process of a delayed fracture of welded joints.

At the first stage of investigations using implant method and method of a composite sample with tension along the weld axis [5], the resistance of HAZ and weld to cold crack formation was studied in accomplishment of a single-layer hardfacing. Both methods provide quantitative evaluation of crack resistance of the above-mentioned regions of the welded joint, and give also opportunity to find indirectly the interrelation between the process of a delayed fracture and level of tensile stresses which initiate it. These studies allowed us to analyze the effect of diffusive hydrogen and heat input of welding on the process of cold crack formation.

All the specimens used during investigations were manufactured from steel 14KhGN2MDAFB ($\sigma_{0.2} =$ = 750 MPa, σ_t = 860 MPa, δ = 20 %, ψ = 65 %) of the following chemical composition, %: 0.15C; 0.23Si; 1.3Mn; 0.97Cr; 2.2Ni; 0.33Mo; 0.41Cu; 0.07Al; 0.015N₂; 0.14V; 0.01S; 0.02P. Welding of specimens was made with solid wires of 1.2 mm diameter in gas

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mixtures Ar + 22 % CO₂ at the following conditions: $I_{\rm w} = 170-180$ A; $U_{\rm a} = 26-27$ V; $v_{\rm w} = 12$ m/h. Specimens, used for evaluation of effect of heat input of welding $q_{\rm w}$, was an exception.

Effect of $[H]_{diff}$ on resistance to a delayed fracture of metal of weld and HAZ was determined in welding with wire Sv-10KhN2GSMFTYu. Hydrogen content in weld was controlled by varying the humidity of CO_2 used. In this case the concentration of $[H]_{diff}$ was changed from 3--4 to 6.5--7.0 ml/100 g. Content of diffusive hydrogen in the deposited metal was evaluated using a chromatographic method [6]. Relationships, characterizing the interrelation between the resistance of welded joint regions investigated to cold crack formation and their saturation with hydrogen, are presented in Figure 1. As it follows from Figure, with increase in content of diffusive-mobile hydrogen the resistance to a delayed fracture of metal both of HAZ and weld is decreased abruptly.

Figure 2 gives the results of investigations of effect of q_w on resistance against the delayed fracture of HAZ metal. The same as in the previous case, the welding of specimens was made with wire Sv-10KhN2GSMFTYu. Concentration of $[H]_{diff}$ in the deposited metal during investigations remained unchanged ---- 6.5-6.8 ml/100 g.

It was established that with increase in q_w from 8.6 up to 20.3 kJ/cm the characteristics of critical stresses σ_{cr} , at which no cracks are occurred in HAZ, are increased from 175 up to 270 MPA. Probably, it







Figure 2. Effect of welding heat input on resistance against delayed fracture of HAZ metal of joints of steel 14KhGN2MDAFB welded by wire Sv-10KhN2GSMFTYu

is due to the fact that with increase in q_w the rate of HAZ cooling within 600--500 °C interval is decreased from 40 to 15 °C/s. Here, the structure of HAZ metal is changed from martensitic (M) to bainitic-martensitic. Moreover, the increase in welding heat input leads to the increase in time of metal cooling in the 800--100 °C interval that contributes to the growth in HAZ metal resistance to cold crack formation.

At the same time q_w can have not only the positive, but also the negative role in the metal resistance to cold cracking. It was established earlier [7] that in joints of similar high-strength steels with BM welds the increase in heat input of welding leads to the increase in residual longitudinal stresses in metal of weld and HAZ. This should be taken into account when selecting the welding conditions.

The effect of weld metal type on resistance to a delayed fracture of metal of welds and HAZ was evaluated from the results of tests of samples welded with wires Sv-10KhN2GSMFTYu (BM weld), Sv-08G2S (FB weld) and Sv-08Kh18N9G7T (AF weld). Chemical and phase composition of weld metal of welded joints made by these wires, and also their characteristics of their strength are given in Table 1. Content of diffusive hydrogen in metal, deposited with these wires, was equal, respectively, to 3.5--4.0, 7.5--8.0 and 2.5--3.0 ml/100 g that is typical of the abovementioned materials.

It was found (Figure 3) that AF of weld metal has a high resistance against the delayed fracture ($\sigma_{cr} / \sigma_{0.2} = 1$). Welds of FB type are characterized by the high resistance against crack formation ($\sigma_{cr} / \sigma_{0.2} = 0.8$). The lower characteristics ($\sigma_{cr} / \sigma_{0.2} = 0.5$) were observed in testing samples with BM welds. Probably, this is associated with peculiarities



Figure 3. Resistance to delayed fracture od metal of welds made by wires Sv-08G2S (1), Sv-08Kh18N9G7T (2), Sv-10KhN2GSMFTYu (3)

of structural transformations in the welds tested, and also with the fact that BM welds have higher values of strength and lower ductility as compared with AF and FB welds (Figure 3).

Structural transformations in BM welds are occurred in the temperature interval at which the metal has the high strength and decreased ductility (Figure 4). In this case, the favourable conditions are created for the occurrence of local plastic deformations in welds and formation of microcracks in them.

In metal of ferritic-pearlitic (FP) welds the structural transformations occur at temperatures above 600 °C. As the strength is low in FP welds at these temperatures, and the elongation exceeds 30 % (Figure 4), then the probability of formation of cracks in these welds is little.

Somewhat another regularities were observed in investigation of HAZ metal (Figure 5). HAZ metal of welded joints made with wires Sv-10KhN2GSMFTYu (375 MPa) and Sv-08Kh18N9G7T (350 MPa) are characterized by the highest characteristics σ_{cr} . The lower values (175 MPa) are obtained in welding with wire Sv-08G2S. The latter is, probably, associated with increased content of diffusive hydrogen in metal deposited with wire Sv-08G2S, and also with the higher (as compared with BM and AF welds) coefficient of hydrogen diffusion in this metal, that contributes to its active accumulation near the fusion boundary at the areas where the cold cracks are formed.

Then, the resistance to formation of longitudinal and transverse cracks of real butt joints with multipass welds was evaluated. For this purpose the technological samples of «rigid contour welding» of different

Table 1. Chemical and phase compositions of weld metal of welded joints made by the wires investigated

Grade of wire				Element	s in weld	metal, wt	.%			Phase	σ _{0.2} ,
Grade of whe	С	Si	Mn	Cr	Ni	Мо	Ti	S	Р	composition, %	MPa
Sv-10KhN2GSMFTYu	0.08	0.29	0.95	0.65	2.1	0.49	0.023	0.023	0.020	70B + 30Ì	780
Sv-08G2S	0.10	0.61	1.49	-	-	-		0.029	0.021	50F + 50B	480
Sv-08Kh18N9G7T	0.09	0.87	5.27	18.7	9.0	-	0.720	0.008	0.026	94À + 6F	350





Figure 4. Dependence of elongation δ_5 (1–3) and reduction in area ψ (4–6) BM (1–5), FP (2, 6) and AF (3, 4) welds on temperature at $w_{6/5} \approx 5$ –6 °C/s

modifications (Figure 6) of a variable width (B = 100, 200 and 300 mm) were used [8]. The samples were manufactured from steel 14KhGN2MDAFB of 15 mm thickness. Their welding was performed without preheating using 1.2 mm diameter solid wires of Sv-08G2S, Sv-10KhN2GSMTYu, Sv-08Kh18N9G7T grades in gas mixture Ar + CO₂ at the following condition: $I_w = 130$ -140 A; $U_a = 24$ -25 V; $v_w = 13$ -14 m/h. Fulfillment of each layer after a root bead was started after cooling of welded joints to 20--30 °C temperature. The process of initiation and propagation of cracks in the welded joints investigated was controlled by the method of acoustic emission [9]. Residual stresses in samples were determined by the method of holographic interferometry [10].

Stressed state of welded joints with multipass welds was evaluated using the samples-satellites. To avoid the possibility of formation of cracks in them, the technological samples were welded with a complete penetration. The highest level (can exceed 500 MPa) of both longitudinal σ_x , and also transverse σ_y residual welding stresses was formed in metal of weld and HAZ of technological samples with B = 100 mm made with wire Sv-10KhN2GSMFTYu (Figure 7, curves 3) unlike the joints welded with wires of Sv-08G2S and Sv-08Kh18N9G7T grades (Figure 7, curves 1, 2) where it is much lower. The values of residual stresses depend greatly on sizes of technological samples: the lowest stresses (from 70 up to 250 MPa) are formed in samples at B = 300 mm [8].



Figure 5. Resistance against delayed fracture of HAZ metal of welded joints of steel 14KhGN2MDAFB made by wires Sv-08G2S (1), Sv-08Kh18N9G7T (2), Sv-10KhN2GSMFTYu (3)



Figure 6. Technological samples for evaluation of resistance of welded joints to longitudinal (a) and transverse (b) cracks: B = 100, 200 and 300 mm; n = 0.5-1.5 mm; n' = 3-4 mm

Resistance of welded joints to the formation of cold cracks was evaluated from the results of investigation of sections cut from the technological samples. Longitudinal cracks at the initial stage of initiation and propagation are spread in the fusion zone of joints, and then transferred into weld metal. This gives grounds to consider that the resistance of welded joints to the formation of longitudinal cracks is defined by the susceptibility of HAZ metal to this type of fracture.

As the investigations showed, the transverse cracks in welded joint made from low-carbon high-strength steels are initiated more often and propagated in weld metal and depend on its ability to resist the delayed fracture. Therefore, the results the investigations of technological samples directed to the evaluation of resistance of welded joints against formation of longitudinal cracks were compared with data obtained during testing implant samples, while at transverse



Figure 7. Distribution of transverse (a) and longitudinal (b) residual stresses in welded joints of steel 14KhGN2MDAFB made by wires Sv-08G2S (1), Sv-08Kh18N9G7T (2), Sv-10KhN2GSMFTYu (3) in weld length L



Figure 8. Ratio of crack height to weld height *C* in length of welded joints*L* of steel 14KhGN2MDAFB made by wires Sv-08G2S (1), Sv-08Kh18N9G7T (2), Sv-10KhN2GSMFTYu (3)

cracks ---- with data obtained at axial tension of weld metal.

The investigations showed that there are no both longitudinal and transverse cracks in welded joints made by wire Sv-08Kh18N9G7T (Figure 8, Table 2). This is confirmed by the results of analysis of macrosections, cut out from samples after 10 days of completion of welding (Figure 9, *a*). As was stated earlier, the residual stresses in technological samples, made by the above-mentioned material, do not exceed 300--320 MPa, that is lower than the level of critical stresses, which can lead to the delayed fracture of metal of HAZ and weld of these joints ($\sigma_{cr} \approx 350$ MPa).

When analyzing the results of testing technological samples and specimens welded by wire Sv-08G2S, it was found that these joints are susceptible to the formation of longitudinal cracks. The first acoustic signals indicating the beginning of the delayed fracture process in welded joints made by wire Sv-08G2S, were recorded 3 min after the completion of welding of sample root weld. After next 5 min the crack of 45 mm length (total length of sample was 300 mm) was observed at the weld surface visually.

During fulfillment of next layers of weld after the root bead the intensity of acoustic signals was greatly decreased. This proves that the crack propagation was delayed further. The crack did not appear at the weld surface. Crack, formed in welding technological samples by wire Sv-08G2S, is shown in Figure 9, *b*. Analysis of macrosections of the welded joint showed that the source of fracture is located at the fusion line from



Figure 9. Macrosections of welded joints of «rigid contour welding» of steel 14KhGN2MDAFB of 15 mm thickness made by wires Sv-08Kh18N9G7T (a) and 08G2S (b) $(\times 3)$

the side of HAZ zone. First, the crack is propagated in HAZ metal, and then it is transferred into weld. Step trajectory of a cold crack in the given weld is oriented to the places of clusters of non-metallic inclusions that is typical of joints with multi-layer welds. Probably, this is explained by the fact that as a result of repeated high-temperature heating, proceeding simultaneously with thermodeformational processes, the decrease in adhesion of non-metallic inclusions with a metal matrix and precipitation of carbides at the grain boundaries is possible, as the sources of cold cracks initiate even at a low degree of hardening at these boundaries [1].

Investigations carried out to evaluate the resistance of welded joints to the formation of transverse cracks showed that the formation of this type of defect in joint of low-carbon high-strength steels welded by wire Sv-08G2S is hardly probable.

The increased susceptibility of welded joints of high-strength steels made by wire Sv-08G2S to the formation of longitudinal cracks and simultaneously high resistance to the formation of transverse cracks can, probably, be explained by different resistance of metal of weld and HAZ of these joints to the delayed fracture under the action of the applied load and also by their different susceptibility to hydrogen embrittlement.

Table 2. Effect of technological sample width and level of residual stresses on resistance to formation of transverse cracks in joints with different type of weld metal alloying

				Width of t	echnological s	ample, mm			
Grade of wire	100			200			300		
	σ _x , MPa	σ _ό , MPa	Ñ, %	σ _x , MPa	σ _ό , MPa	Ñ, %	σ _x , MPa	σ _ό , MPa	Ñ, %
Sv-08Kh18N9G7T	309	200	0			Not do	torminod		
Sv-08G2S	356	350	0		Not determined				
Sv-10KhN2GSMFTYu	550	350	70	350	200	50	245	160	0
Note. Here C is the ratio	of crack heig	ht to weld he	ight.						

Analysis of macro- and microsections made from technological samples welded by wire Sv-10KhN2GSMFTYu showed that both longitudinal and transverse cracks can be formed in them. It was found by investigations that at a limited content of diffusive hydrogen in the deposited metal it is possible to prevent the formation of longitudinal cracks in joints welded by welding wire Sv-10KhN2GSMFTYu when the level of residual stresses in them does not exceed 350 MPa. This is confirmed by the results of testing technological samples of 200 and 300 mm, in which the maximum levels od residual stresses are equal, respectively, to 340 and 220 MPa. In case, when the level of stresses exceeds 350 MPa (technological sample of 100 mm width), it is not possible to prevent the crack initiation in specimens without additional technological procedures (use of preliminary and auxiliary heating). These data have a good correlation with results of testing implant samples which proves that at comparable welding conditions $([H]_{diff} = 3.5 -= 4.0 \text{ ml}/100 \text{ g}, q_w = 8 - 10 \text{ kJ/cm})$ the initiation and propagation of cold cracks in HAZ metal of the joints investigated can occur at loads of more than 375 MPa.

Results of testing the technological samples made by wire of Sv-10KhN2GSMFTYu grade to evaluate the resistance of joints to transverse crack formation prove also an important role of residual stresses in the cold crack formation.

It was found that the high resistance of welded joints to this type of fracture can be attained at the condition that the level of residual longitudinal stresses in them will not exceed 240 MPa (Table 2). It should be noted that these data have a good correlation with the results of tests for a delayed fracture of weld metal ($\sigma_{cr} = 240$ --260 MPa). In practice, to reduce the longitudinal stresses to the above-mentioned level, different technological procedures are used, among them the cascade, uphill welding, dividing of long areas of welded into separate blocks of a small length etc. should be noted.

CONCLUSIONS

1. It was established that with increase in concentration of diffusive hydrogen the resistance of metal of weld and HAZ of welded joints of high-strength steels to a delayed fracture is decreased.

2. It is shown that increase in heat input of welding promotes, in the one hand, the formation of structural

constituents in HAZ metal of high-strength hardening steel welded joints, that is characterized by an increased resistance to a delayed fracture, and, from the other hand, increase in level of residual tensile stresses.

3. It was found that increase of residual stresses in welded joints made from low-carbon alloy highstrength steels causes the decrease in their resistance to cold crack formation both in longitudinal and transverse directions relative to the weld axis.

4. It was established that the rigid fixture and type of weld metal influence greatly the formation of residual stresses in high-strength steel welded joints. The lowest level of residual tensile stresses was observed in rigidly-fixed joints of steel 14KhGN2MDAFB with austenitic and ferritic-bainitic welds (350--375 MPa), and the highest level (up to 600 MPa) ---- with bainitic-martensitic welds.

5. Design of welded joints and selection of technological processes of high-strength steel welding should be done so that to prevent in them the feasibility of formation of high level of residual welding stresses, imperfections of microstructure of metal of weld and HAZ, and also the presence of areas with an increased concentration of the diffusive hydrogen.

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CHARACTER OF FORMATION OF HOT CRACKS IN WELDING CAST HEAT-RESISTANT NICKEL ALLOYS

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Mechanisms of formation of hot cracks in HAZ during fusion welding have been studied. It is presumed that heating of the HAZ metal of a heat-resistant alloy during welding causes structural transformations of the strengthening γ -phase, accompanied by changes in metal volume. Hot cracks are induced by the deformation processes localised in a narrow zone, where the nickel alloy structure remains fully austenitic for a certain period of time.

Keywords: heat-resistant nickel alloys, γ -phase, heat-affected zone, hot cracks, hardness, plastic deformation, ductilitydip range

Heat-resistant nickel alloys are the main structural materials used to manufacture gas-turbine engines widely applied in aircraft industry and other engineering sectors.

Components of a hot section of engines are made from strain-hardening nickel alloys reinforced by intermetallic compounds. To ensure stability of structure and maintain high long-time strength properties of nickel alloys, they are subjected to complex alloying with γ -forming elements (aluminium, titanium, niobium), the total content of which in an alloy is 6-15 wt.% or more.

The complex alloying system and high strength characteristics of the alloys are responsible for formation of cracks both in weld and HAZ in welding parts of even insignificant thickness. So, it is appropriate to study mechanisms of formation of hot cracks first of all in HAZ of the joints made by fusion welding accompanied by structural transformations.

The studies were conducted on nickel alloy IN 738, which is a structural material used to manufacture gas turbine blades. The alloy has the following chemical composition, wt.%: 0.09C, 16.0Cr, 10.5Co,

1.7Mo, 4.6W, 0.2Nb, 3.0Al, 4.4Ti, $Ca \leq 0.01$ and La $\leq 0.01.$

Sensitivity of the HAZ metal to cracking in plasma-powder welding was evaluated on the basis of a composition identical to that of the base metal. Welding was performed on the IN 738 sample 10 mm thick in the as-received state under the following conditions: $I_{\rm W} = 100$ -120 A, $U_{\rm a} = 25$ -26 V, $v_{\rm W} = 4$ m/h, powder particle size ---- 50--150 μ m, and argon flow rate ---- 18--20 l/min.

Sections for metallographic examinations were subjected to vacuum ion etching using a high-voltage plasma discharge at a voltage of 2.5 kV and current of 0.005 A, which made it possible to clearly detect cracks of a different size. Optical microscope lightand dark-field images were examined to prove the presence of the zones of structural transformations revealed in the HAZ metal.

Sizes of structural components of metal, including the γ' -phase, were determined by scanning electron microscopy after special etching. The character of plastic deformation in propagation of a hot crack was evaluated from variations in profile of the welded joint surface by interference optical microscopy.

Microcracks propagating in a direction of base metal (Figure 1) were detected in the HAZ metal by



Figure 1. Microcracks in HAZ metal of welded joints in nickel alloys containing the strengthening γ -phase (×200)



Figure 2. Light- (a) and dark-field (b) images of microcracks in HAZ metal of a welded joint in heat-resistant alloy IN 738 (\times 100)

OURNAL



Figure 3. Variations in width L_z of the zone of full $\gamma \rightarrow \gamma \rightarrow \gamma'$ transformation in welding a heat-resistant nickel alloy containing the strengthening γ -phase at a different initial temperature: a - 77; b - 293; c - 1273 K (×50)

metallography of welded joints in alloy IN 738, conducted at normal temperature of the weld edges. Examination of these welded joints revealed the presence of a structural zone adjoining the weld, in which the hot cracks were predominantly concentrated (Figure 2).

Statistical processing of the examination results showed that the cracks were located at some distance from the fusion line. This suggests that it is not always melting of grain boundaries during welding that causes formation of hot cracks. Therefore, these cracks cannot be classed with typical solidification cracks.

Zones with differing properties are formed in base metal near the fusion line under the effect of the welding thermal cycle. Study of the mechanism of formation of these zones will help to reveal the nature of hot cracking.

In general, structural changes in the HAZ metal of welded joints depend upon the welding conditions and, primarily, upon the entire welding thermal cycle.

Studies were conducted to investigate the effect of initial temperatures of samples welded on formation of the concerned structural zone, the welding conditions being left unchanged.

Welding was performed at room temperature (293 K) on metal preliminarily cooled to 77 K and heated to 1293 K.

Microstructure of the HAZ metal of welded joints in alloy IN 738 is shown in Figure 3. It can be seen from the Figure that width L_z of the zone of full γ -transformation undergoes substantial changes. Exponential curves of temperature dependence of L_z (Figure 4) were plotted on the basis of the data obtained. As proved by statistics, width L_z of the zone of full γ -transformation correlates to a mean length of the hot cracks formed in welding of metal with a different initial temperature. Examination of microstructure of metal of the zone of interest made it possible to reveal particles of the finely dispersed γ' -phase present in the zone adjoining the weld (Figure 5, *a*, *b*). These particles differ in size from coarser particles of the γ' -phase in base metal, which is formed at a distance from the fusion line (Figure 5, *c*).

Scanning electron microscopy confirmed differences in structure and size of the γ' -phase particles in the base metal zone located at a distance from the weld (Figure 6, *a*) and in the hot cracking zone adjoining the weld (Figure 6, *b*).

Therefore, width L_z of the zone of full γ -transformation with an increased hardness, as well as that of the hot cracking zone, depends upon the initial temperature conditions of welding. Here, the zone of an increased hardness has a minimal width in the welded







Figure 5. Microstructure of HAZ metal in heat-resistant nickel alloy IN 738: a — general view (×50); b, c — fragments 1 and 2, respectively (×500)

joint made on metal preliminarily cooled to 77 K. The above differences can be explained by the effect of the rate of metal cooling within a temperature range of $\gamma \rightarrow \gamma'$ transformation on the diameter of the strengthening γ' -phase particles.

It is shown in [1] that increase in cooling rate v_{cool} is accompanied by decrease in the diameter of the γ -phase particles (Figure 8). Here one might expect changes in strength characteristics of the welded joint metal, including hardness.



Figure 6. Size of particles of the strengthening γ -phase in different zones adjoining the weld: $a - d = 0.1-0.9 \mu m$ (base metal zone); $b - d = 0.005-0.150 \mu m$ (hot cracking zone)

In a general form, strength of the strain-hardening alloys depends upon the distance between the particles, their diameter and volume fraction [2], and is determined from the following formula:

$$\sigma = \sigma_0 + \frac{c\sqrt{f}}{d},$$

where σ_0 is the tensile strength of the matrix; *c* is the constant that includes the Burgers vector and shear modulus of the matrix; and *f* is the volume fraction of structural components.

It follows from the equation that the larger the volume fraction of the strengthening γ' -phase and the smaller the diameter of the particles, the stronger the alloy.

Therefore, it can be concluded that metal of the hot cracking zone adjoining the weld has an increased hardness (strength) due to the $\gamma' \rightarrow \gamma \rightarrow \gamma'$ transformation occurring under the effect of the welding thermal cycle (heating \rightarrow cooling) (Figure 9). Here differences in hardness are related to variations in the rate of metal cooling in welding and, accordingly, the $\gamma \rightarrow \gamma'$ transformation process taking place in cooling of the HAZ metal after the weld pool has moved in



Figure 7. Dependence of microhardness of HAZ metal upon its initial temperature: *1* — cooling to 77 K; *2* — room temperature (293 K); *3* — heating to 1273 K





Figure 8. Changes in mean diameter *d* of the γ -phase particles in cooling from temperature $T > T_{f.d}$ (here *T* is the initial temperature, and $T_{f.d}$ is the temperature of full dissolution of the γ -phase particles): O — data of study [1]

a direction of the welding line. Schematic of the preferential formation of cracks in HAZ of the welded joints in nickel alloys with the γ -phase strengthening and its relationship to structural changes in the HAZ are shown in Figure 10.

Propagation of a hot crack along the grain boundaries from the γ -phase zone, where there are no particles of the strengthening γ -phase (Figure 10), in a direction of the base metal is decelerated due to relaxation of local stresses induced by plastic strains at the crack mouth (Figure 11, *a*). As seen from Figure 11, *b*, here we have the plane strain condition. It is likely that in this case the mechanisms of deceleration of a crack can be explained on the basis of conditions of formation of plastic strain zones described by fracture mechanics for metallic materials. The results obtained are proved by optical interference metallography of the mouth surface of the hot crack formed in nickel alloy after welding. Figure 12 shows the pattern of distribution of interference fringes at the crack mouth and in a region of artificial strain induced by a scratch on the section surface. It can be seen that the interference fringes in the crack mouth and scratch regions tend to move upwards. As the scratch makes a recess in metal, the zone of metal ahead of the crack moves downwards relative to the section plane, i.e. the metal shrinks due to realisation of the plastic strain mechanism.

Results of the studies suggest that the hot cracks in HAZ of nickel alloys strengthened by the γ -phase are formed in welding by the following mechanism:







Figure 10. Schematic of preferential formation of cracks in HAZ of welded joints in heat-resistant alloys containing the strengthening γ -phase



Figure 11. Deceleration of hot crack in HAZ under the effect of plastic strain: a — metal surface at the hot crack mouth (×400); b — schematic of distribution of plastic strain in stress fields (σ_1 - σ_4)



SCIENTIFIC AND TECHNICAL



Figure 12. Distribution of plastic strains at the hot crack mouth

1. Heating of the HAZ metal in welding of an alloy causes structural transformations of the strengthening γ -phase into the γ -matrix. These transformations occur in a temperature range from about T = 700 °C to $T_{\rm f.d.}$ depending upon the selected system for metal alloying.

2. The $\gamma \rightarrow \gamma'$ transformation is accompanied by increase in metal volume, as at increased temperatures the crystalline lattice parameters of the γ -phase are lower compared with those of the γ -phase. This is accompanied by development of intensive thermal-deformation processes localised in a narrow zone of HAZ.

3. There is a high-temperature zone adjoining the weld, wherein structure of a nickel alloy remains for some time in a fully austenitic state.

In this case the austenite grain boundaries are enriched with impurity elements, i.e. carbon, oxygen, sulphur, phosphorus and other surface-active elements. The effect of plastic strain on the embrittlement processes is realised through interaction of mobile dislocations with impurity atoms, and determined by competing temperature processes, such as variations in energy of interaction of atoms with dislocations (inverse temperature dependence) and diffusivity of these atoms (direct temperature dependence).

The temperature range, in which a sufficiently high energy of binding between an impurity atom and dislocations persists and, at the same time, mobility of this atom is provided due to increase in the diffusion parameters, is the ductility-dip temperature range. It is in this range that an impurity element can be captured by moving dislocations and transported to the grain boundaries.

The latter circumstance leads to a change in the grain body to boundary strength ratio and formation of a crack along the grain boundaries (by the type of ductility-dip cracks). As noted earlier, embrittlement of the grain boundaries takes place in the zone with an austenitic structure. As the metal cools down, the γ -phase is precipitated inside the grains. This is accompanied by development and, in the case of increase in volume fraction of the γ -phase, enhancement of the processes of deceleration of dislocations and enrichment of the boundaries with impurity elements, as well as increase in ductility of the HAZ metal. Further propagation of cracks is stopped in the said range because of plastic strain at the crack mouth.

CONCLUSIONS

1. Nickel alloys strengthened by the γ -phase are typically characterised by formation of cracks along the grain boundaries in the HAZ during heating-cooling within a range from 700 °C to temperatures close to the melting point.

2. Phase transformations $\gamma + \gamma' \rightarrow \gamma \rightarrow \gamma + \gamma'$ and local deformation are proved to occur along the grain boundaries within the base metal zone close to the fusion line during welding. Relaxation of stresses takes place in a temperature range from 700 °C to the melting point in a crack formed at the grain junction (at the crack mouth).

3. Hot cracks may be formed under the effect of thermal-deformation loading as a results of segregation of impurity elements at the grain boundaries, followed by their melting and substantial local deformation in the ductility-dip temperature range.

4. In metals characterised by a high degree of alloying with surface-active elements, e.g. boron, the cracking process may occur simultaneously by two mechanisms.

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INFLUENCE OF WELDING SPEED AND DURATION OF PERIODICAL COOLING ON STRUCTURE FORMATION IN WELDED JOINTS OF HARDENING STEELS IN ARC WELDING WITH THERMAL CYCLING

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Proceeding from analysis of the relationship between the heating and cooling conditions of welded joints it is shown that the main welding parameters upon which the efficiency of thermal cycling depends, are welding speed and pause duration. These parameters determine the heating rate, number of heating and cooling cycles and thermal cycling interval, which, in turn, determine the kinetics of structural transformations in heating and cooling of a welded joint.

Keywords: arc welding, hardening steels, thermal power, thermal cycling, structure

Structure of welded joint metal is an important factor, determining its properties. This dependence is particularly evident in welding of hardening steels. Therefore, study of the features of the influence of various factors on structure formation in welding of the above group of steels is an important and urgent task, as it allows development of new advanced methods of structure improvement.

It is known that conditions of metal heating and cooling have a predominant influence on the kinetics of metal structure formation. Features of this influence with the traditional welding processes have been rather profoundly and comprehensively studied [1--5]. With the traditional welding processes, the heat input determines the heating and cooling conditions. Reduction of this input is accompanied by accelerated heating and cooling of welded joints, thus promoting a limitation of the growth of the austenitic grain and higher susceptibility of welded joints to hardening. At increased heat input the rate of welded joint heating and cooling becomes lower. This results in increased grain growth and different degrees of martensite self-tempering.

With the conventional processes of arc welding, it is difficult to ensure a simultaneous refinement of the grain and lower probability of hardening structure formation. This is rather labour-consuming, expensive and not practicable. For this reason a number of techniques have been recently developed, which enable welding of hardening steels without preheating. One of such techniques is thermal cycling.

Work [6] gives the results of dilatometric studies, which indicate that thermal cycling allows shifting austenite decomposition to higher temperature regions and suppressing martensite transformation in the welded joint. In this case, the metal develops structures which are the result of austenite decomposition in the intermediate or diffusion regions. This is achievable at pulsed-arc single-pass welding without preheating or heat treatment of the welded joint due to initiation of a specific pulsed thermal cycle in it of the type of that given in [6]. Increase of heating rates, and number of heating--cooling cycles and expansion of the temperature interval of thermal cycling promote shifting of austenite decomposition to higher temperature regions and formation of more ductile structures in the welded joint.

The main parameters of a pulsed thermal cycle that determine the kinetics of structure formation in a welded joint, are the heating rate, number of heating--cooling cycles, thermal cycling temperature range and cooling rate. As the influence of cooling rates on structure formation has been rather profoundly studied, there is no point in considering it. Each of the above thermal cycle parameters has a corresponding parameter of the welding mode, which has a predominating influence on it.

The heating rate is mainly determined by the welding heat input, which, in its turn, is controlled by welding speed at a fixed value of thermal power [7, 8]. Therefore, the heating rate at a fixed thermal power of the arc in the pulse is mainly determined by welding speed in the pulse, and rises with its increase. Temperature range of thermal cycling is determined by the pause duration. Number of heating--cooling cycles depends on two parameters, namely welding speed in the pulse and pulse duration. With minimum pulse duration the number of heating--cooling cycles can be increased at different speeds of welding in the pulse.

Minimum pulse duration depends on the inertia of the feed mechanism and, as shown by practical experience, it varies in the range of 0.5--1.5 s, depending on the design of feed mechanism. In welding with thermal cycling it is rational to maintain the minimum pulse duration. At a fixed duration of the pulse it may be assumed that the number of heating-cooling cycles is determined by the speed of welding in the pulse.

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Figure 1. Influence of welding speed in the pulse on phase composition and microhardness of the structural components in the HAZ metal of welded joints on steel M72 made by welding with thermal cycling

The main purpose of this work was studying the influence of these parameters on the kinetics of formation of welded joint structure. Investigations were conducted on welded samples of steel M72 (up to 0.78 % C) up to 20 mm thick. Welding was performed with 65G wire of 1.6 mm diameter. The pulse current varied from 150 to 400 A, and pulse voltage from 24 to 34 V.

Investigations were performed in two stages. In the first stage the speed of welding in the pulse was measured at a fixed thermal power of the arc in the pulse. The weld pool was periodically brought to complete solidification. In the second stage the pause duration was varied at fixed values of the other parameters of the mode. Microhardness of the structural components was determined by $HV_{0,2}$.

Results of the first stage of investigations are given in Figures 1--4. As is seen from Figure 1, at welding speeds in the pulse between 9.8 and 30 m/h the HAZ



Figure 2. Martensite-troostite structure in the HAZ metal of welded joint on steel M72 made by welding with thermal cycling (×240)



Figure 3. Bainite structure of HAZ metal of welded joints of steel M72 made by welding with thermal cycling (×400)

metal forms a structure, which consists of martensite and troostite (Figure 2). Presence of troostite component in the structure is indicative of a sufficiently high level of inhomogeneity of part of the austenite and lowering of its overcooling stability. This results in its decomposition in the temperature field of diffusion transformation. Troostite microhardness varies, being close to HV 300.

Part of austenite has enough time to homogenize. It results in an increase of its stability at overcooling, whereas the decomposition shifts into the region of diffusionless transformation with martensite formation. Martensite microhardness gradually decreases with increase of welding speed. If at $v_{w.p} = 9.8 \text{ m/h}$ the microhardness is HV 593, $v_{w.p}$ increase up to 19.4 m/h is accompanied by microhardness lowering to HV 566, and at $v_{w.p} = 26 \text{ m/h} HV$ 513. This is indicative of a gradual increase of the temperature of martensite transformation. Therefore, at increase of welding speed from 9.8 to 26.0 m/h the level of austenite homogeneity gradually drops and its decomposition shifts to higher temperature regions.

In the speed range of 30--35 m/h conditions are in place, when all of the austenite has a sufficiently high level of inhomogeneity and low stability at overcooling. This is sufficient to initiate its decomposition in the intermediate region with bainite formation (Figure 3). With increase of welding speed in the pulse from 30.3 to 35.2 m/h, bainite microhardness drops from HV 441--460 to 396--412. Therefore, temperatures of the start and finish of austenite decomposition gradually rise, and the slight difference between the maximum and minimum microhardness confirms the fact that the difference between these temperatures is also small, i.e. austenite has an approximately same level of inhomogeneity, which gradually increases with welding speed rising from 30.3 to 35.2 m/h.

Further increase of the welding speed above 35.2 m/h is negative for the kinetics of structure formation in the HAZ metal. This is confirmed by the martensite-troostite structure, in which the amount of martensite is increased with increase of welding speed, and the difference between the minimum and maximum microhardness of the structural components

becomes markedly greater. At $v_{w.p} = 39.5 \text{ m/h}$ the maximum microhardness (of martensite) is HV 700, and minimum (of troostite) is HV 303. Increase of welding speed in the pulse up to 47 m/h is accompanied by increase of maximum microhardness up to HV 793, and of minimum value up to HV 386. Therefore, at increase of welding speed above 35.2 m/h the level of austenite decomposition gradually shifts to the low-temperature regions, which even do not have the conditions for development of self-tempering of the products of diffusionless decomposition. This is due to the features of the thermal cycle of welding.

In welding with thermal cycling the kinetics of structure formation depends on a combination of the heating rates, number of heating--cooling cycles and temperature range of thermal cycling. The higher the heating rates, the greater the number of heating--cooling cycles and wider the thermal cycling range, the higher are the temperatures of the start and finish of austenite decomposition. With increase of welding speed in the pulse, the heating rates are increased, but the number of heating--cooling cycles is decreased. For each specific value of the arc thermal power there exist the threshold speeds of welding in the pulse, at which the number of heating--cooling cycles is so small, that they practically do not have any influence on the kinetics of structure formation. Therefore, after such a threshold has been exceeded, austenite decomposition can shift into the region of lower temperatures with formation of a martensite component.

The diagram in Figure 4 allows evaluation of the speed of welding in the pulse on grain size in the HAZ metal. It is seen that at increase of the speed of welding in the pulse from 9.8 to 26.0 m/h the secondary austenite grain grows from 5--6 to 3--4 points, respectively. In the range of speeds of 26.0--39.5 m/h the grain dimensions are stabilized within 3--4 points. Further increase of the speed of welding in the pulse is accompanied by a gradual refining of the grain to 5--6 points at $v_{w.p} = 47$ m/h.

Influence of welding speed in the pulse on the structure of weld metal (Figure 5) is similar to that in the HAZ metal. With increase of welding speed in the pulse to 30.3 m/h, the difference between the maximum and minimum microhardness of the structural components in the weld metal decreases. At $v_{w,p} = 30.3 \text{ m/h}$ the metal structure is gradually transformed into a purely bainite structure with microhardness *HV* 412 and remains to be such up to $v_{w,p} = 35.2 \text{ m/h}$, being, however, more heterogeneous as to its microhardness. Further increase of the welding speed is accompanied by a greater difference between the minimum and maximum microhardness of the structure and appearance of martensite in it.

These investigation results correlate quite well with the data of work [7], which describes the features of the influence of welding speed on the thermal cycle at thermal cycling. According to this work, with other mode parameters being constant, increase of welding speed to a certain value promotes higher heating rates

Figure 4. Influence of welding speed in the pulse on grain point in HAZ metal of welded joints on steel M72 made by welding with thermal cycling

and manifestation of a pulsed mode of the thermal cycle.

The first stage of investigations showed that improvement of the conditions of formation of welded joint structure by variation of just the welding speed is of a limited nature. Despite the fact that suppression of martensite transformation in the weld and HAZ metal can be achieved here, any significant refinement of the grain is highly problematic.

Further improvement of the welded joint structure, as shown by the second stage of investigations, can be achieved through further widening of the temperature interval of thermal cycling, which is achieved by increasing the pause duration.

In the previous stage of investigations, the most acceptable results were obtained at the speed of welding in the pulse of 30.3 m/h. Therefore, the second

Figure 5. Influence of the speed of welding in the pulse on phase composition and microhardness of structural components in the weld metal of welded joints on steel M72 made by welding with thermal cycling by 65G wire

Figure 6. Influence of pause duration on phase composition and microhardness of structural components of welded joints on steel M72 made by welding with thermal cycling with 65G wire: a — weld; b — HAZ metal

stage was conducted at its fixed value. Results of these investigations are shown in Figures 6--8.

As is seen from Figure 6, *a*, at pause duration of up to 1.7 s, the weld develops a bainite-martensite structure with a sufficiently great difference between the maximum and minimum microhardness of the structural components. This difference is gradually becoming smaller at greater duration of the pause. At pauses of 1.7--2.6 s the weld metal forms a bainite structure, where the microhardness does not exceed HV 460, and at pause duration longer than 3.5 s, the microhardness drops to HV 300 and lower. Further prolongation of the pause to 3.8 s and more is accompanied by formation of a mixed bainite-troostite-pearlite-ferrite structure in the weld metal with the microhardness HV 230--260, which decreases as the pause becomes longer.

Figure 7. Bainite-troostite-pearlite-ferrite structure in the HAZ metal of welded joints on steel M72 made by welding with thermal cycling (×400)

The above regularity is valid also for HAZ metal (Figure 6, b). Maximum microhardness is found in a structure which is a mixture of martensite and troostite, which forms at pause duration of 0.8 s. In this case the maximum microhardness (martensite) is higher than HV 600, and the minimum value (troostite) is not more than HV 260. This is indicative of the fact that part of austenite had the level of homogeneity, which led to its decomposition in the diffusion region. Part of austenite with a higher level of homogeneity and higher resistance at overcooling decomposed in the martensite region, and, judging from microhardness, at lower temperatures. Martensitetroostite structure in the HAZ metal forms right up to the pause duration of 3.2 s, except for 1.7 s pause, at which the HAZ metal develops a bainite structure with microhardness HV 441--460. Therefore, at increase of the pause duration, martensite microhardness gradually drops, i.e. martensite transformation shifts to higher temperature regions and self-tempering of the transformation products develops.

Increase of the pause duration to 3.2-3.5 s provides formation of a bainite-troostite structure in the HAZ metal with a microhardness varying in the range of HV 400 (bainite) and HV 300 (troostite), while gradually becoming lower. At pause duration of more than 3.5 s the HAZ metal develops a bainite-troostite-pearlite-ferrite structure (Figure 7). Its microhardness is HV 286--382 and gradually drops to HV 276--312 at a pause of 4.1 s.

A longer pause provides a gradual shifting of austenite decomposition to higher temperature regions. In this case the maximum and minimum microhardness values of the structural components are gradually becoming smaller, and so does the difference between them.

It should be noted that increase of the pause duration is accompanied by refinement of the secondary austenite grain (Figure 8). For instance, at pause duration of 1.7 s, the grain point is 4-5 to GOST 5639-82. Prolongation of the pause to 3.5 s is accompanied by grain refinement to 6-10 points, with prevailing of finer grains. At pause duration of 4.1 s

IRNAL

Welding process	Factor of increase of penetration depth	δ, <i>mm</i>	I, A	U, V	v _w , m∕h	q, J/cm
Unactivated		10	550	40-42	25	24964
Activated	1.82.0	10	400	34	40	9189
Same	2.02.5	10	300	30	30	8103
»	3.04.0	10	220	25	36	4400

Arc welding heat input at different penetration depths

the grains are refined to 7--11 points, the main bulk of the grains corresponding to 9--11 points.

Data of the second stage of investigations are indicative of the fact that prolongation of the pause and widening of the temperature interval of thermal cycling approximately to 1400--450 °C leads to lowering of the level of austenite homogeneity and its stability at overcooling. As a result, austenite decomposition gradually shifts to higher temperature regions, thus providing formation of softer and more ductile structures in the welded joint. Widening of the temperature interval of thermal cycling also results in refinement of the secondary austenite grain.

Thus, structures ranging from the martensite to bainite-pearlite-troostite mixtures can be formed in the welded joint at the same welding speed and thermal power of the arc in the pulse by changing the pause duration and controlling the temperature interval of thermal cycling. It should taken into account that pause duration, required for shifting austenite decomposition into the regions of the intermediate and diffusion transformations, can be quite large. The greater the heat input, the longer should be the pause, and this impairs the welding efficiency. Therefore, the current goal is to shorten the pause with simultaneous increase of the intensity of periodical cooling of the welded joint. One of the ways to solve this problem is lowering of the welding heat input, as a shorter pause is required for cooling to a certain temperature at a lower heat input. With the conventional approach to solving this problem the welding efficiency will also decrease.

Activation of the arc allows solving the problem of lowering the heat input without detracting from the welding efficiency. The Table gives the results of experiments to evaluate the influence of penetration depth in consumable electrode welding. Activation (through increasing the arc penetrability) allows lowering the heat input, compared to the traditional welding processes. As is seen from the Table, in welding metal of the same thickness arc activation lowers the welding heat input by 2.7 to 6.7 times.

CONCLUSIONS

1. Increase of the welding speed in the pulse at fixed values of the other parameters of the mode and periodical solidification of the weld pool allows prevention of martensite transformation in the section of the HAZ metal overheating.

2. Variation of the pause duration in the range of 0.8 to 4.1 s at unchanged welding speed allows adjustment of the temperature interval of thermal cycling and formation of a structure ranging from the martensite to bainite-pearlite-martensite mixtures in the welded joint.

3. Compared to the conventional welding processes, arc activation allows lowering the heat input at welding metal of the same thicknesses and increasing the intensity of periodical cooling of the welded joint in the pause.

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43

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STRUCTURE AND PROPERTIES OF THIN-SHEET EB WELDED JOINTS IN IRON--NICKEL ALLOY 32NKA

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Data on electron beam weldability of iron-nickel alloy 32NKA of the type of Invar are presented. Parameters for welding different types of the joints for the fabrication of thin-walled shell structures have been optimised. Chemical composition, mechanical properties, microstructure, density, hardness and chemical heterogeneity of welded joints have been investigated. X-ray diffraction analysis of the recrystallisation zone has been conducted. Recommendations on the use of electron beam welding for the commercial fabrication of welded structures are given.

Keywords: electron beam welding, weldability, electron beam, nickel alloy, mechanical properties, microstructure, thin-sheet shell

Advances in engineering require new approaches to the fabrication of structures of special materials, allowing implementation of different design and technology solutions, which cannot be realised by traditional methods. For example, precision nickel alloys with special thermal properties and preset linear thermal expansion coefficients (LTEC) of about 10^{-6} K⁻¹ and lower have found application in metrology, cryogenic and space engineering [1, 2].

The values of LTEC close to zero are needed to ensure high precision of measuring tools, as well as casings and some units of optical systems used in gas lasers and telescopes. Recent developments in the field of telescopes have been stipulated by their use at space facilities to survey the state of the earth surface from space to handle problems in different types of construction work and in geodesy. Important peculiarities of operation with these devices in space are special temperature conditions from 100--120 °C (under increased solar radiation) to --100 -- --120 °C (where a facility is in the shadow of the earth), and deep natural vacuum.

To ensure optical and operational properties of the equipment, and allowing for the peculiarities

Figure 1. Temperature dependence of linear thermal expansion coefficient for iron-nickel alloys: 1 - 64.8; 2 - 57.7; 3 - 46.3; 4 - 42.0; 5 - 35.4 % Ni

of the space conditions, the Fe--Ni alloy of the 32NKA grade, having a minimum LTEC of about $\alpha \leq 3.5 \cdot 10^{-6}~K^{-1}$, was recommended as a structural material to study weldability.

LTEC of alloys Fe--Ni--Co with a nickel content below 60 wt.% has an abnormal character. The level of minimum expansion corresponds to an alloy containing 36 % Ni. Alloys with this or close nickel content are called Invars. Table 1 gives chemical composition of some grades of alloys of the Invar type.

The so-called «Invar effect» shows up in alloys with a nickel content of about 36 % (plus or minus, i.e. in a range of 30-40 %). Alloys of this group are characterised by a strong dependence of the Curie temperature and LTEC upon the concentration of their basic elements. In addition, it is necessary to allow for the effect the elements exert on the temperature range of the Invar behaviour determined by the Curie temperature in heating and $\alpha \rightarrow \gamma$ transition temperature in cooling to room temperatures or below. This characteristic can be especially pronounced in the recrystallisation zone during the process of joining by fusion welding.

Joining of alloys Fe--Ni--Co can be performed by argon-arc, microplasma, laser and incoherent light beam welding, as well as by electron beam welding in vacuum [3--8].

The method of EBW in vacuum was chosen for the studies in view of the above-said and considering that the use of small-thickness materials for large-size structures may cause serious difficulties with the

Table 1. Chemical composition of iron--nickel alloys

Alloy	Content of elements, wt.%								
grade	Ni	Со	Fe	С	Si	S	Р	Mn	
33N	36		Base	0.05	0.3	0.012	0.02	0.5	
35NKT	34.5	5.5	Same	0.05	0.5	0.015	0.02	0.3	
32NK-VI	32.5	4.2	»	0.02	0.03	0.015	0.02	0.2	

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Figure 2. Types of EB welded joints in alloy 32NKA: a — butt joint; b — overlap joint; c — T-joint; 1, 2 — members joined; 3 — welds

manufacture precision and prevention of distortions in making vacuum-tight welds with a narrow recrystallisation zone to maintain desired LTEC [9].

Optimal parameters of the process and conditions for EBW of alloy 32NKA were selected using specimens 0.8 mm thick. Figure 2 shows variants of the resulting types of the joints, and Table 2 gives welding parameters.

Welding was performed by using the U-250A power unit with an accelerating voltage of 30 kV.

Figure 3. Macrostructure of welded joints: a — butt joint; b — overlap joint; c — T-joint

Figure 4. X-ray pattern of EB welded joints in alloy 32NKA 0.8 mm thick: a — butt joint; b — overlap joint with through penetration

Figure 5. Distribution of microhardness in different zones of welded joints in iron-nickel alloy 32NKA: a — butt joint; b — T-joint

Visual examination of the joints and analysis of transverse macrosections (Figure 3) showed full and through penetration in butt and overlap joints, and absence of through penetration in T-joints, but on both sides and with an overlap in the weld root, which complied with the requirements for welding telescope tube structures. Continuity of welded joints was controlled by the X-ray methods using the RUP unit. No pores or other defects were detected in butt and overlap joints (Figure 4). Distribution of microhardness was examined using the LECO hardness meter. Indentations were made under a load of 20 g and holding time of 15 s, the distance between them was 200 μ m (Figure 5). Metallography was performed using the «Neophot-32» optical microscope.

Chemical composition of base and weld metals was determined by spectral analysis, the results of which are given in Table 3.

 Table 2.
 Welding parameters used to make joints in alloy 32NKA

Joint type	Welding current, mA	Focusing current, mA	Welding speed, m/h	
Butt joints	18	72	50	
Overlap joints	20, 30, 35	72	50	
T-joints	22	72	50	

Table 3. Chemical composition of regions of the EB welded joint in alloy 32NKA

Joint region	Content of elements, wt.%					
Joint Tegton	Ni	Со	Fe			
As-received base metal	31.9	3.95	Base			
Recrystallisation zone	31.8	4.0	Same			
Weld metal	31.5	3.9	»			

Figure 6. Microstructure of base metal and recrystallisation zone of the EB welded joint in alloy 32NKA: a ---- base metal; b — fusion zone; c ---- recrystallisation zone of the HAZ metal in its central part (×150)

Figure 7. Microstructure of the EB welded joint in alloy 32NKA in the recrystallisation zone to weld metal transition region: a ---recrystallisation zone near the weld (×150); b -- fusion zone (×150); c -- weld metal (×250)

Chemical microheterogeneity of the joints was studied by the X-ray microanalysis methods using the CAMECA microanalyser SX-50. According to the Xray microanalysis results, the distribution of nickel and cobalt in the joints was uniform (Table 3).

Mechanical ultimate strength tests were conducted on butt specimens. As shown by the test results, the value of ultimate strength was 525 MPa, which corresponded to about 74 % of that of the base metal in the heat-treated condition.

Microstructure and chemical microheterogeneity of welded joints were studied on the butt joint specimens. This was done using the «Neophot-32» optical microscope and CAMEBAX microanalyser with microscope SX-50. The studies showed grains deformed by rolling (Figure 6, *a*) in the base metal, changing into equiaxed grains in the recrystallisation zone (Figure 6, *b*), as well as disappearance of fibrous grains and formation of a microgranular structure with thickened grain boundaries in this region. The grains increased in size with distance to the weld (Figure 6, *c*), acquiring a polyhedral shape, and grain boundaries becoming thinner. Transition from the recrystallisation to weld zone was uniform (Figure 7, *a--c*).

Table 4. Local chemical composition of regions of welded joint in alloy 32NKA determined by using the CAMEBAX X-ray microanalyser

Joint region	Content of elements, wt.%						
Joint region	Si	Mn	Со	Ni	Fe		
Base metal	0.213	0.001	4.168	31.982	63.636		
Recrystallisation zone	0.226	0.001	4.252	31.875	63.646		
Weld metal	0.224		4.360	31.906	63.510		

X-ray microanalysis (both by the point-by-point method with a locality of 1 μ m (Table 4) and by scanning (Figure 8) of all the regions) revealed a high degree of chemical homogeneity of metal of the joints: in all the cases the content of nickel was 31.5--31.9 %, cobalt ---- approximately 4.0--4.2 %, the balance being iron. Manganese in the form of traces was detected at the fusion boundary. A small amount of silicon (up to 0.25 %), being an associated element, was fixed, its distribution in the joints also being uniform.

Therefore, the microstructures observed were caused not by chemical heterogeneity, but by structural changes in the recrystallisation zone, as well as by a probable $\gamma \rightarrow \alpha$ transformation, which may lead to a substantial decrease in the «Invar effect» and density of the alloy in the recrystallisation zone. In turn, this leads to increase in LTEC and loss of vacuum tightness in thin-walled structures.

X-ray diffraction analysis was conducted in K_{α} cobalt radiation using the DRON-3m unit to study phase and structural transformations in the recrystal-

Figure 8. Distribution of nickel (1) and cobalt (2) alloying elements in the recrystallisation zone, transition zone and weld metal of a butt joint in alloy 32NKA

Table 5. Phase composition of welded joint in alloy 32NKA in the recrystallisation zone

Dif	fraction cha	Alloy phase composition		
20, deg	θ, deg	d_{α} , nm	J	
51.25	25.625	0.20701	100	γ-Fe
52.80	26.40	0.20131	41	γ-(Fe, Ni, Mn, Co)
53.50	26.75	0.19887	15	γ-(Fe, Ni)
53.75	26.875	0.19805	12	β — weak line
59.50	29.75	0.18039	12	γ -(Fe, Ni), β -line
59.90	29.95	0.17929	52	γ-(Fe, Ni)
74.90	37.45	0.14721	30	Not detected
89.80	44.90	0.12681	47	γ -Fe + γ -(Fe, Ni)
100.3	50.15	0.11659	36	α-Fe
110.8	55.40	0.10874	12	$\gamma Fe,~\beta-$ weak line
111.7	55.85	0.10816	32	γ- Fe

lisation zone of a welded joint. The spectrogram obtained (Figure 9) shows phase composition of the alloy comprising the γ -(Fe, Ni) based solid solution. It is likely that manganese and cobalt were dissolved in this solution, which can be seen from the changed values of interfacial distances d_{α} (Table 5). No lines of pure nickel or cobalt were detected. Single lines of a small amount of α -Fe were revealed. This proves the «Invar effect» taking place in this region, which was cooled from the melting point to room temperature. Therefore, the chosen method of EBW characterised by high speeds with a minimal heat input, narrow HAZ and minimal section of the weld can be

Figure 9. Spectrogram of phase composition of recrystallisation zone of the EB welded butt joint in alloy 32NKA

considered optimal. So, it can be recommended for the commercial manufacture of telescope components and tubes.

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47

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GENERALIZATION OF FORMULA OF K.K. KHRENOV FOR DETERMINATION OF TEMPERATURE OF WELDING ARC PLASMA

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Formulae to determine temperature of the welding arc plasma as a function of pressure in the plasma have been derived using the Steenbeck principle of the arc minimum. It is shown that the Khrenov's formula becomes precise at a very high pressure, while at low pressure the formula, defined more precisely, is valid, giving the decrease in temperature and increase in degree of ionisation of plasma with pressure reduction in the arc plasma.

Keywords: welding arc, temperature of plasma, pressure in plasma, calculation

In 1949 the article of Prof. K.K. Khrenov was published [1], in which he was the first who derived the formula for determination of temperature of electric arc plasma, containing only universal physical constants, using the principle of Steenbeck arc minimum [2]. Khrenov's formula was included to all the manuals because of its simplicity. However, the solution, obtained by him, does not take into account the effect of pressure in plasma on temperature of arc plasma due to use of assumptions.

Results, obtained in the article, are valid not only for conventional arcs at atmospheric pressure, but also for arcs of a decreased pressure $(10^2 - 10^4 \text{ Pa})$ used in different technologies of spraying and deposition of coatings, and also for arcs of high pressure of $10^7 - 10^8 \text{ Pa}$ order, observed in underwater welding and also in high-pressure arc tubes.

Saha's equation for equilibrium plasma in a full form [3] at a single ionisation of metal vapours and gases in arc column has the form

$$\frac{\alpha(T, p)^2}{1 - \alpha(T, p)^2} = \left(\frac{2\pi m_e}{h^2}\right)^{3/2} \frac{a^2}{p} (kT)^{5/2} \exp\left(-\frac{eU_i}{kT}\right) \quad (1)$$

where α is the degree of plasma ionisation; *T* is the temperature of plasma, K; *p* is the pressure in plasma, Pa; m_e is the mass of electron equal to $9.1095 \cdot 10^{-31}$ kg; *h* is the Plank's constant equal to $6.626176 \cdot 10^{-34}$, J·s; $a = \sqrt{2g_i/g_a}$; g_i/g_a are the ratio of static weights of ions and atoms, different for different metals (see Table; more detailed tables of these values can be find in works [4, 5]); *k* is the Boltzmann's constant equal to $1.3806 \cdot 10^{-23}$ J/K; U_i is the potential of the first ionisation of metal vapours, V; *e* is the electron charge equal to $1.6022 \cdot 10^{-19}$ C.

Let us consider the channel model of an arc cylindrical column of radius R_0 . Arc column is in a thermal equilibrium state. Heat, generated in arc, is dissipated from the surface by the arc radiation. At electric field intensity E in arc column and arc current I the equa-

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tion of power balance for unit of the arc column length can be written in the form

$$IE = 2\pi R_0 \beta \sigma_{\rm B} T^4, \qquad (2)$$

where β is the degree of plasma emissivity equal to 0.55--0.65; σ_B is the Stefan-Boltzmann constant equal to 5.6687·10⁻⁸ W/ (m²·K⁴). At the same time, for the channel model of arc at current density *j*, constant in arc cross-section,

$$I = j\pi R_0^2. \tag{3}$$

Here, the current density j is mainly stipulated by a current density constituent equal to [4]

$$j = en_e b_e E, \tag{4}$$

where n_e is the concentration of electrons in plasma; b_e is the mobility of electrons equal to $b_e = (ae\lambda_e) / m_e v_e$; for electrons a = 1; $v_e = \sqrt{3kT/m_e}$ is the root-mean-square heat rate of electrons.

In expression (4) the length of a mean free path of electrons for high-pressure arcs at low degrees of ionisation α is taken usually equal to $\lambda_e = 1/ng_{ea}$, where *n* is the initial concentration of non-ionised vapours of metals; g_{ea} is the section of collision of electrons with neutral atoms (Ramsauer section). For metals, used in welding, the values g_{ea} are within the wide range (28--300)·10⁻²⁰ m² (approximate values g_{ea} are given in the Table). It is impossible to make allowance for collision of electrons with ions for lowpressure arcs, when the degree of ionization is very high. Therefore, the formula for the mean free path length should be written in the form

$$\lambda_e = \frac{1}{n_a g_{ea} + n_i g_{ei}} = \frac{1}{n g_{ea} (1 + \alpha k_g)}.$$
 (5)

Here $n_a = n(1 - \alpha)$ is the concentration of neutral molecules of metal vapours in plasma; $n_i = \alpha n$ is the concentration of ions of vapours of metal in plasma; g_{ei} is the cross-section of collision of electrons with ions; coefficient $k_g = g_{ei} / (g_{ea} - 1)$.

Theory of effective cross-sections of collision of electrons with ions [3, 6] does not yet provide the

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Dependence of temperature (a) and degree of ionisation (b) of plasma of arc, burning in vapours of different metals, on pressure (channel model)

satisfactory correlation with experimental data. In existing procedures of interaction of bombarding electron with ion is considered as purely coulomb interaction even with allowance for electron screening. Actually, in the presence of electron screening this interaction is little differed from interaction of electron with a neutral atom, the number of valency electrons of which is decreased by a unit (single ionisation) or several units (multiple ionisation). According to data of work [3] the Ramsauer section g_{ea} for different atoms is increased approximately 2 times with decrease in valency electrons by unit. Therefore, it is possible to consider that the cross-section of collision of electrons with ions is increased by the same number of times: at single ionisation $g_{ei}/g_{ea} = 2$. As the number of multiple ionised atoms, as compared with a number of single ionised atoms in the low-temperature plasma is small, then in calculation of coefficient k_{α} it is possible to take value $g_{ei}/g_{ea} = 2$, and $k_g = 1$. This value of coefficient k_g will give the best correlation with experimental data [7--12].

Excluding R_0 from equation (2) and (3) we shall find

$$\left(\frac{Ie^2}{4\pi\beta^2\sigma_{\rm B}^2g_{ea}\sqrt{3km_e}}\right)E^3 = T^{17/2}\left(k_g + \frac{1}{\alpha(T, p)}\right)$$
(6)

Principle of arc minimum formulated by Steenbeck [2] consists in the fact that at fixed I and p the arc tries to provide the minimum possible value of electric

Parameters of some chemical elements

field intensity *E* in arc column, at which the requirement $\partial E / \partial T = 0$ is fulfilled.

By differentiating the right part of the equation (6) by *T* at p = const and equating the derivative to zero, we shall obtain the following equation after transformation at $k_g = 1$:

$$\left(\frac{eU_i}{kT} + \frac{5}{2}\right) [1 - \alpha(T, p)] = 17.$$
 (7)

Formula (7) allows us to express p as function from T in a clear form. At $\alpha \rightarrow 0$ we shall come to the known formula of K.K. Khrenov [1, 13] in which the plasma temperature does not depend on pressure:

$$T = 2eU_i/29k \approx 800U_i. \tag{8}$$

49

The obtained relationships T(p) for different metals are given in Figure. Pressure p means the total pressure in plasma consisting of pressure of medium p_0 and excessive electromagnetic pressure in arc p_{em} , stipulated by the electromagnetic forces. Values U_i and a for different metals used in calculations are given in the Table.

After determination of T(p) it is possible to obtain the dependencies of coefficient of ionisation α on the pressure for plasma of vapours of different metals (Figure *b*). Minimum (boundary) pressures, at which the obtained curves are still valid, are defined by the conditions of conservation of quasi-thermicity of arc column plasma (and conservation of arc column it-

Parameter	С	Al	K	Ti	Cr	Fe	Zr	W
U_i , V	11.26	7.8	4.34	6.8	6.74	7.87	6.92	7.98
à — coefficient	1.155	0.577	1	1.633	1.309	1.549	1.633	1.549
g_{ea} 10 ²⁰ , m ²	50	100	300	50	35	35	50	35

BRIEF INFORMATION

self). These conditions are different for different metals, the region of boundary pressures has a rather unclear nature and is within the 5--100 Pa interval.

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EFFECT OF ELECTRODE STICKOUT ON WELD PARAMETERS IN PULSED-ARC WELDING OF STEELS

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Effect of changes in electrode stickout on geometric parameters of welds in pulsed-arc consumable electrode welding was studied. It was found that application of system of automatic stabilization of mean values of arc voltage with an effect on parameters of power source pulses does not allow compensation of change in weld width. System of automatic stabilization of mean values of arc voltage and welding current with a proper effect on parameters of pulses of power source and electrode wire feed speed prevents the decrease in weld width caused by increase in electrode stickout.

Keywords: pulsed-arc welding, consumable electrode, electrode stickout, weld parameters, systems of automatic stabilization

Application of different algorithms for control of parameters of the process of a pulsed-arc consumable electrode welding (PACEW) under the action of disturbing factors was stipulated by the need in improvement of the weld quality. At the E.O. Paton Electric Welding Institute the new principles of design of systems of automatic stabilization of the PACEW process were developed using the feedback circuits for stabilizing the mean values of arc voltage U_a^{av} and welding current I_w^{av} [1, 2].

The aim of the present work was to study the effect of electrode stickout on weld parameters and effectiveness of application of the developed systems of automatic stabilization. Two systems of automatic stabilization were used: U_a^{av} with action on parameters of power source pulses, and also system of automatic stabilization U_a^{av} and I_w^{av} with action, respectively, on parameters of power source pulses and electrode wire feed speed.

Deposits were made on inclined plates from 16 mm thick steel of 14G2 grade with 1.2 mm diameter wire Sv-08G2S in mixture of shielding gases Ar + + 18 % CO₂ at linear increase of stickout L_s . Initial parameters of welding at $L_s = 13$ mm are given in the Table.

The Figure shows the relationships of changing the geometric parameters of deposited welds (weld width B, penetration depth H, height of convexity a) with increase in electrode stickout at different initial values of welding current and welding speed. The Figure shows that with increase in electrode stick-

Dependencies of weld width (a), penetration depth (b), convexity height (c) on changes in electrode stickout L_s at different values of welding current: 100 (1), 190 (2) and 250 (3) A (\bigcirc — without system of stabilization; \square — with system of stabilization of U_a^{av} ; \triangle — with system of automatic stabilization of U_a^{av} and \vec{I}_w^{av})

8/2004

BRIEF INFORMATION

Weld- ing cur- rent, A	Arc voltage, V	Wire feed speed, m∕h	Welding speed, m∕h	Gas consump- tion, 1/min	Pulse duration, ×10 ⁻³ , s	Pulse frequen- cy, Hz
100	21.8	207	20	14	3.2	44
190	25.2	405	23	15	3.2	87
250	26.7	532	27	17	3.2	105

out the weld width and penetration depth are decreased and the weld connexity is increased. These data are increased with increase in welding current.

Application of the system of automatic stabilization of U_a^{av} makes it possible to compensate the change in penetration depth and convexity height. However, the weld width is changed in this case. This can be explained by the peculiarities of effect of I_w^{av} in PACEW of steels on weld width, where the change in I_w^{av} has the greater influence on weld width than on penetration depth and convexity height. And, as $I_{\rm w}^{\rm av}$ is not stabilized in use of the system of stabilization of $U_{\rm a}^{\rm av}$, then the weld appearance is deteriorated due to decrease in its width. The latter is manifested greater at increased welding currents, as with increase in electrode stickout the higher oscillations of $I_{\rm w}^{\rm av}$ are occurred.

Analysis of relationships, given in Figure, allows us to conclude that the change in electrode wire stickout in PACEW of steels leads to the deterioration of geometric parameters of welds. The application of system of automatic stabilization of U_a^{av} with action of power source pulses on parameters does not compensate the changes in weld width occurring here. The use of a two-circuit system of automatic stabilization of U_a^{av} and I_w^{av} makes it possible to maintain geometric sizes of welds more precisely at variations of the electrode stickout.

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