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# WELDED FREIGHT BOGIE BOLSTER PROJECT

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The possibility of making welded freight bogie bolsters to compete with traditional cast ones is considered. It is shown that making of a welded structure, for which the cyclic load plays a decisive role, requires an integrated approach based on development of optimal designs of units and manufacturing technologies.

*Keywords: welding, welded joints, bolster, fatigue life, stress-strain state, stress concentration, oxidising shielding gases, welding parameters, heat input, mechanical properties, microstructure, non-metallic inclusions, acicular ferrite*

Freight bogies applied currently in the CIS countries use bolsters made from steel castings. Production of cast bolsters is characterised by a high probability of internal defects and, hence, inconsistency of strength values, particularly under cyclic loading. Until recently, Ukraine has had no home production of bolsters, despite a considerable volume of production of freight rolling stock, the equipment for which has been bought from abroad. In this connection, the «Azovmash» Company made an attempt to manufacture a bolster made by welding.

Production of bolsters of new designs for railway rolling stock is regulated by corresponding standards [1] providing for a certain scope and procedure of tests. To design weldments, the emphasis should be given to cyclic loading resistance, which is a factor that determines weight and, therefore, economic indicators of a bolster. In terms of cyclic loading, the regulations provide for two stages of the tests: curtailed accelerated fatigue tests and, in a case of positive results, comprehensive fatigue tests.

Accelerated bend tests under alternating loading per cycle: $P_{max}$ = 800 kN and $P_{min}$ = 100 kN, are used at the stage of development of a new design of bolsters.

Allowable cyclic fatigue life of a bolster, $[N]$, is assumed to be equal to $1 \cdot 10^6$ cycles, i.e. it should exceed 1 mln loading cycles in accelerated fatigue life tests conducted up to formation of a macrocrack 10–50 mm long.

Central Specialised Car-Building Design Bureau of the Joint Stock Company «Azovmash» developed a version of the welded structure of a bolster made from low-alloy steel of the 09G2S grade (Fi-gure 1).

Analysis of the stress-strain state of the bolster performed by the finite element method revealed the presence of local zones with an increased stress level

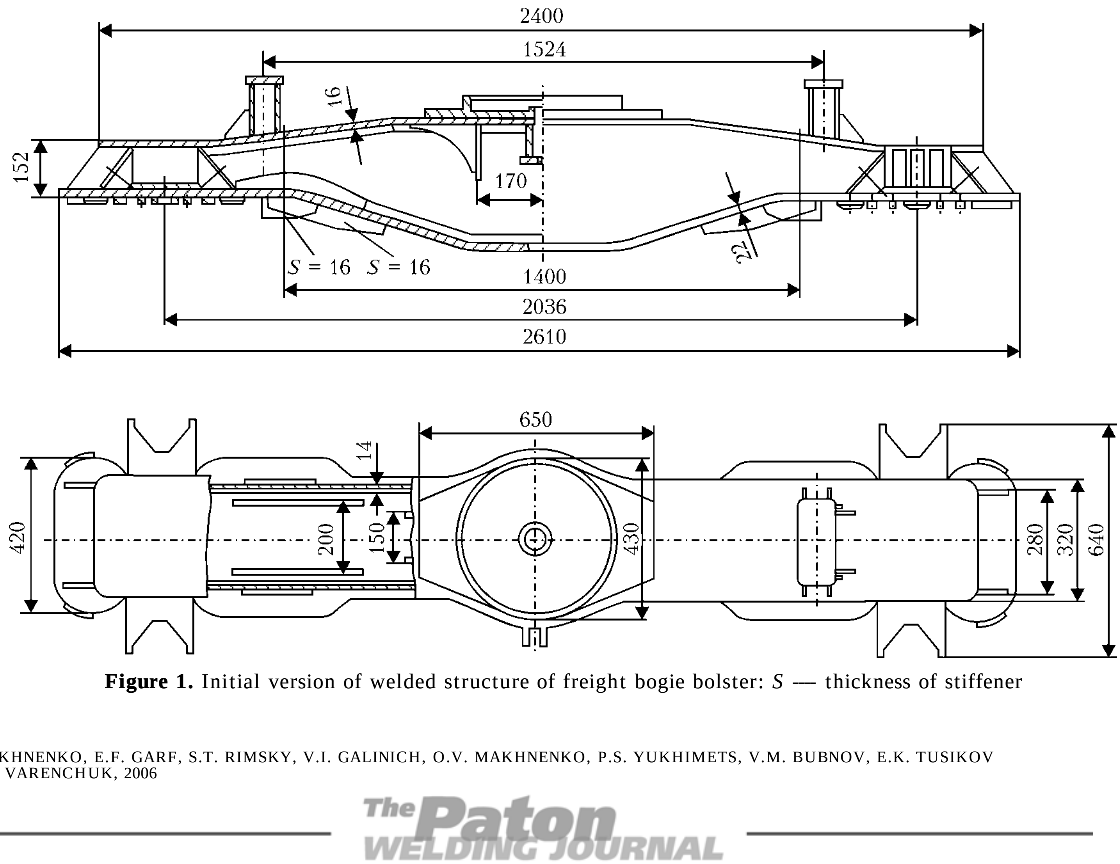


**Figure 1.** Initial version of welded structure of freight bogie bolster: $S$ — thickness of stiffener

(Figure 2). As seen from the Figure, these zones are found mainly in locations of lower chord bends at the end of the bolster. To reduce the level of stresses in these locations, the bolster wall was reinforced with an extra cover plate. In general, the bolster had a number of design and technology drawbacks resulting in a high stress concentration and decrease of its fatigue resistance, the most important of them being the presence of cover plates, stiffeners, technological backing elements, etc.

Accelerated cyclic tests of bolsters made under this project were conducted by the Test Centre «Azovmashtest». Initiation of the first macrocrack was fixed at $N$ = 443,000 loading cycles. The crack was formed at the end of a stiffener welded to the lower chord. In further tests the centres of initiation of new cracks were locations of the ends of stiffeners and cover plates. Totally, seven centres of initiation of fatigue cracks were revealed during the tests. The bolster fractured at $N$ = 735,000 loading cycles.

The level of cyclic stresses corresponding to the actual fatigue life to initiation of fatigue macrocracks was determined by calculations, based on designs of bolster weldments and this fatigue life.

Given that the fatigue calculations are based primarily on nominal stresses, Figure 3 shows diagrams of bending moments and, accordingly, stresses in the lower (tensioned) and upper (compressed) chords of the bolster. The diagrams of stresses give in brackets the values of maximal stresses $\sigma_{max}$ allowing for the cycle asymmetry factor, and without brackets — the amplitudes of cyclic stresses $2\sigma_a$.

Results of the fatigue calculations made for a bolster according to [2] show that with the given bolster design the maximal values of stresses $\sigma_{max}$ on a base of $1\cdot10^6$ loading cycles are equal to 110.1 MPa, and those on a base of 443,000 loading cycles (corresponding to initiation of a fatigue crack) are equal to 130.3 MPa. As the calculations a priori should provide a certain level of safety factor for strength of a structure, it can be noted that correctness of the calculation procedure in the given example [2] is questionable, as the values of actual stresses in the cyclic fracture zone were not in excess of 121 MPa.

Therefore, the fatigue calculations for this design of the bolster were also made by using other methods practiced in machine building and construction industry, e.g. the method developed by the Engineering Science Institute of the Russian Academy of Sciences and the E.O. Paton Electric Welding Institute of the National Academy of Sciences of Ukraine [3–5].

The results show that the fatigue life for the given design of the bolster at $N = 1\cdot10^6$ loading cycles can be provided in the case where the values of stresses are not in excess of 71.1 MPa, while at $N = 4.43\cdot10^5$ loading cycles the value of stresses is 92 MPa.

Application of the IIW recommendations for fatigue design [6] yields the following results: $\sigma_{max}$ = 71.7 MPa at $N = 1\cdot10^6$ cycles, and 94 MPa at $N$ = 443,000 loading cycles.

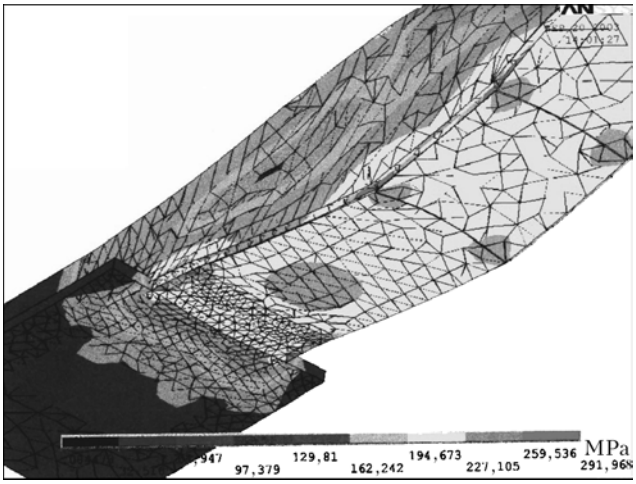


**Figure 2.** Distribution of stresses in bolster determined by the finite element method

Apparently, the results obtained from the last two methods are very close to each other, and they agree with the test results, allowing for a difference between calculated and ultimate stresses for a given fatigue life. Therefore, it is these methods that are recommended for use to develop the bolster design. In addition, as shown by the test results, fatigue cracks were found in the zone of bolster sections *II–II* and *III–III* (see Figure 3), where the level of tensile cyclic stresses is much lower compared with sections *IV–IV*. This is another proof of the fact that the level of stresses is not the only factor that affects the fatigue resistance. A not less effect on the fatigue life of a part may be exerted by its design.

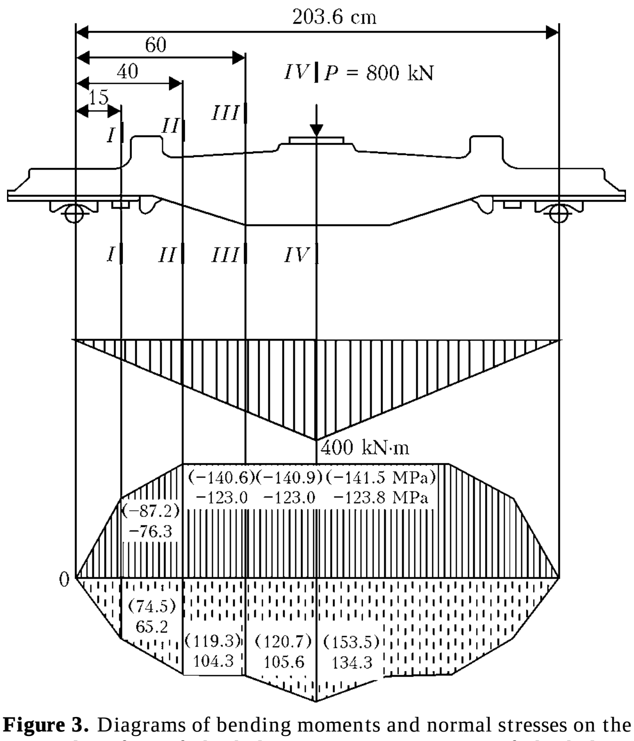


**Figure 3.** Diagrams of bending moments and normal stresses on the external surface of the bolster: *I–IV* — sections of the bolster regions (the rest of the explanations are given in the text)

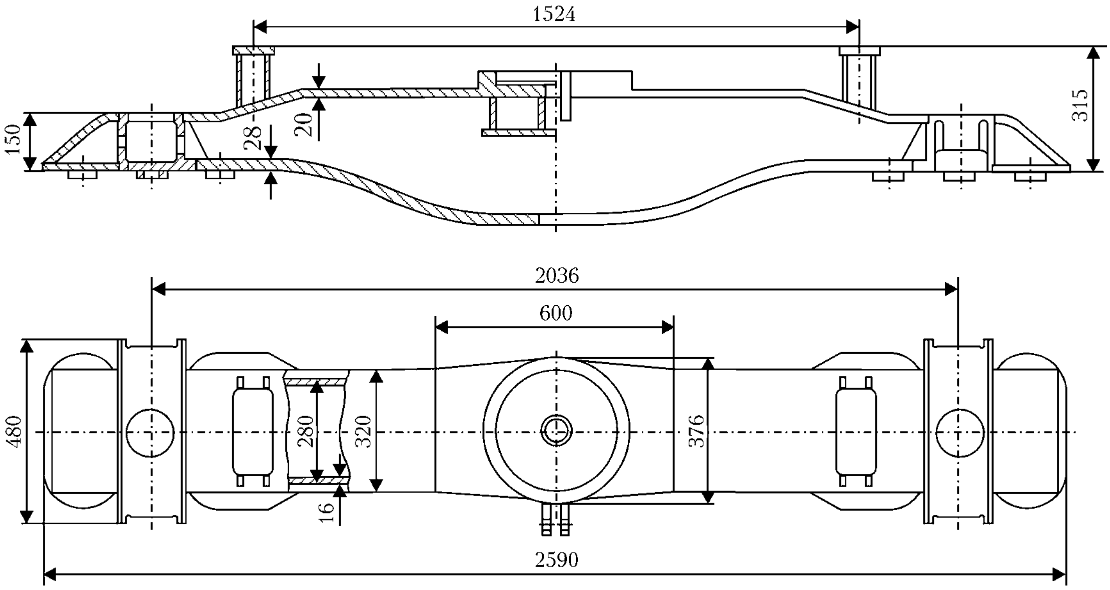


**Figure 4.** Recommended version of a welded structure for freight bogie bolsters

Therefore, to develop a welded bolster having the economically viable properties and required reliability level, it is necessary to attempt, first of all, to use designs that provide a high fatigue resistance. To address this problem, it is necessary to do the following:

• design of the bolster should be made as simple as possible by avoiding extra elements, and cross section of the welds and volume of the deposited metal should be decreased;

• stress concentration in the bolster structure should be decreased, and special attention should be given to elements and assemblies with a high level of tensile stresses. In this connection, stiffeners and cover plates should be eliminated, and butt and fillet welds should be made with full penetration. For the butt joints on tensioned elements, it is expedient to remove the weld reinforcement and eliminate the use of cover plates by providing full penetration and back weld formation by means of special techniques. The fillet welds in the tensioned zone of a bolster should have a concave shape. It is necessary to avoid undercuts in the transverse tensioned butt joints;

• 100 % non-destructive testing by physical methods should be carried out for the most loaded butt welded joints;

• maximal tensile stresses in welded joint zones on a bolster induced by design load should not exceed design resistance to cyclic loading on a base of $N = 1\cdot10^6$ loading cycles. In this case, the fatigue resistance should be estimated by traditional procedures.

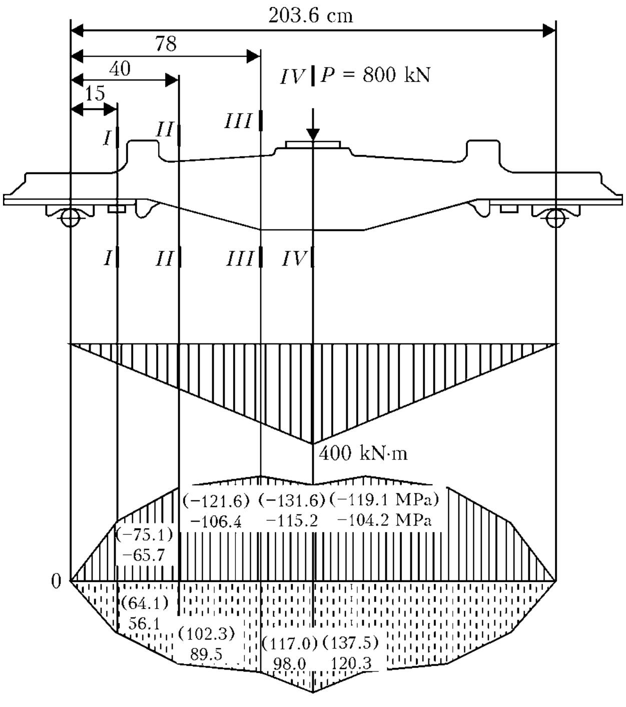


**Figure 5.** Diagrams of bending moments and normal stresses on the external surface of the recommended version of a bolster

The above requirements are embodied to a substantial degree in Project 1690.00.010-6SB worked out by the Design Bureau of the Company «Azovmash» (Figure 4).

Distribution of stresses in tensioned and compressed chords of a bolster is shown in Figure 5. The amplitudes of stresses and maximal stresses in a cycle were decreased by 7–15 % due to increase in cross section of the chords and walls of the bolster. At the same time, the calculated resistance values were considerably increased owing to improvement of the bolster design. In this case, characteristic welded joints are butt joints in chords and joints between the chords and stiffeners. The maximal values of stresses on a base of $N = 1\cdot10^6$ loading cycles for these types of the joints, as determined by the procedures of the Engineering Science Institute of the Russian Academy of Sciences and the E.O. Paton Electric Welding Institute of the NAS of Ukraine, are 153 and 288 MPa, respectively (143 and 180 MPa, respectively, according to the IIW recommendations). Apparently, the above values of stresses are higher than those found in a welded bolster. This guarantees the required level of the fatigue life.

At the same time, the developed bolster design can be practically implemented only by using a high manu-

**Table 1.** Chemical composition of welding consumables and weld metal, wt.%

| Item of analysis | C | Si | Mn | S | P | Cr | Ni | Cu | [O] | [N] | [H] |
|---|---|---|---|---|---|---|---|---|---|---|---|
| Base metal — steel 09G2S 20 mm thick | 0.091 | 0.67 | 1.61 | 0.024 | 0.021 | 0.05 | 0.05 | 0.050 | – | – | – |
| 1.2 mm electrode wire Sv-08G2S | 0.120 | 0.90 | 2.06 | 0.020 | 0.019 | 0.09 | 0.04 | 0.020 | – | – | – |
| Weld metal: | | | | | | | | | | | |
| $CO_2$ welding | 0.095 | 0.50 | 1.40 | 0.022 | 0.020 | 0.08 | 0.04 | 0.035 | 0.067 | 0.011 | 0.00013 |
| Ar + 25 % $CO_2$ welding | 0.084 | 0.58 | 1.45 | 0.024 | 0.021 | 0.11 | 0.06 | 0.050 | 0.032 | 0.010 | 0.00016 |

facturing technology, and the assembly-welding operations in particular. In this connection, the special consideration was given to the development of the technology for welding of bolsters.

The welding process was chosen on the basis of the following requirements: railway rolling stock is operated within a wide range of temperatures, dynamic loads and other unfavourable factors, and optimal combination of mechanical properties of the weld metal is determined by its microstructure [7, 8]. The latter greatly depends upon the metal chemistry, as well as such technological factors as thermal and deformational welding cycles, groove shape, type of a welded joint, etc.

It is a known fact that one of the possible ways of improving the GMAW process for steel is the use of mixtures of shielding gases, instead of $CO_2$ gas, along with such efficient approaches as the application of flux-cored and activated wires for welding, as well as power supplies and welding equipment with characteristics that provide control of melting and transfer of the electrode metal. Many technological problems can be solved by varying the composition of a gas atmosphere. In terms of a combination of high welding-operational and economical indicators, the most promising one is a mixture of argon with oxidising shielding gases ($O_2$, $CO_2$) [6, 8–10].

As shown by the efforts on development of the technology and procedure for welding of bolsters, the optimal mixture in terms of requirements for a shielding gas is Ar + 25 % $CO_2$. Technological advantages of this mixture are particularly pronounced in a range of welding parameters that provide a spray metal transfer. In this case, the surface of the welds was small-scaled and looked like that of the submerged-arc welds. Spattering and splashing of electrode metal occurred 3–4 times less intensive than in $CO_2$ welding.

Welding of low-alloy structural steels in oxidising shielding gases ($CO_2$ and Ar + $CO_2$ mixture) is usually performed using electrode wires Sv-08G2S and Sv-08GS alloyed with silicon and manganese (according to GOST 2246–70). However, despite the fact that they are allowed for use by many regulatory documents, the weld metal produced by using these wires does not always have the required values of impact toughness at negative temperatures and required level of crack resistance [9]. Therefore, it is necessary to determine the process parameters for welding of bolsters to provide the required level of cold resistance of the welds when using standard welding wires.

Cold resistance of the weld metal on low-alloy steel 09G2S (according to GOST 5520–78) was evaluated from the ductile-brittle transition temperature corresponding to the level of impact toughness equal to 47 J / cm$^2$ on sharp-notched specimens (of the type of XI according to GOST 6996–66). The specimens were cut from the upper part of the butt welded joints, and notches were made in a cross section of the weld from the root to top.

Welding was performed with V-groove using 1.2 mm electrode wire Sv-08G2S in a shielding atmosphere of $CO_2$ and Ar + 25 % $CO_2$ mixture under the following conditions: $I_w$ = 280–300 A, $U_a$ = 28–30 V, shielding gas flow rate — 16–18 l / min, and electrode extension $l_e$ = 18–20 mm. The welding speed was varied from 15 to 50 cm / min, which corresponded to a heat input of 10–35 kJ / cm.

Chemical composition of the weld metal is given in Table 1, and mechanical properties are given in Table 2.

Variations in impact toughness at different test temperatures versus welding parameters are shown in Figures 6–8. The results obtained allow evaluation of the parameters which make a trend to increase in cold resistance of the weld metal most pronounced.

Firstly, the composition of a shielding gas atmosphere plays an important role in ensuring high values of cold resistance of the weld metal. When welding is performed in the Ar + $CO_2$ mixture, the ductile-brittle transition temperature is 20–30 °C lower than in $CO_2$ welding (see Figures 6–8). This improvement of cold resistance is provided by a decreased content of oxygen (see Table 1). In turn, this leads to decrease in the volume content of non-metallic inclusions in

**Table 2.** Mechanical properties of metal of the welds made on steel 09G2S using 1.2 mm wire Sv-08G2S

| Shielding gas | $\sigma_{0.2}$, MPa | $\sigma_t$, MPa | $\delta_5$, % | $\psi$, % |
|---|---|---|---|---|
| $CO_2$ | 397–469 / 430 | 535–600 / 577 | 26.3–30.2 / 27.2 | 60.5–68.0 / 67.0 |
| Ar + 25 % $CO_2$ | 473–485 / 480 | 590–630 / 610 | 27.0–29.5 / 28.4 | 59.9–63.2 / 62.8 |

*Note.* Denominator gives average values over the results of testing 3–5 specimens.

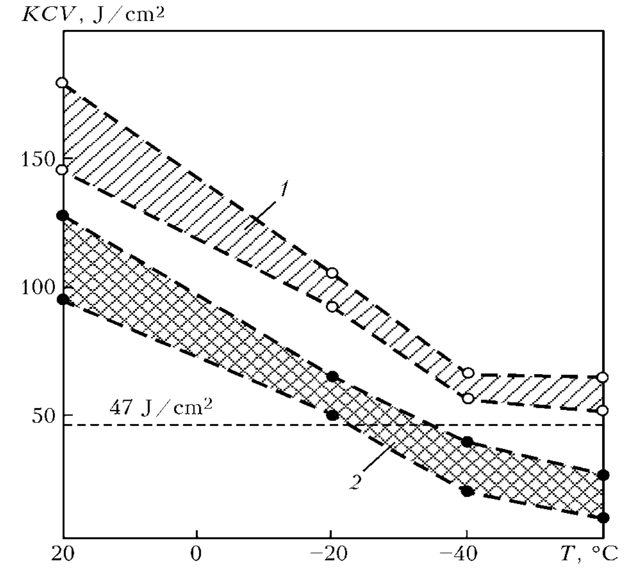


**Figure 6.** Temperature dependence of impact toughness *KCV* of metal of the welds made in Ar + 25 % $CO_2$ mixture (*1*) and $CO_2$ (*2*) on steel 09G2S at 19 kJ / cm heat input

the weld metal and a more uniform distribution of fine inclusions [8, 10, 11] acting as centres of nucleation and epitaxial growth of ferrite in $\gamma \rightarrow \alpha$ transformation, caused by cooling of metal in a temperature range of 800--500 °C and formation of final acicular ferrite structure in it [12].

Secondly, decreasing a heat input of the welding process is a very efficient way of increasing the values of impact toughness of the weld metal, which can be achieved through increasing the welding speed, the rest of the welding parameters being left unchanged. In general, the productivity of welding operations hardly decreases, only the number of welding passes increases, which is caused by decrease in cross section of each of them. Thermal cycle of the weld metal is favourable for the formation of structures of the type of acicular ferrite, as this leads to increase in the rate of cooling in a temperature range of 800–500 °C.

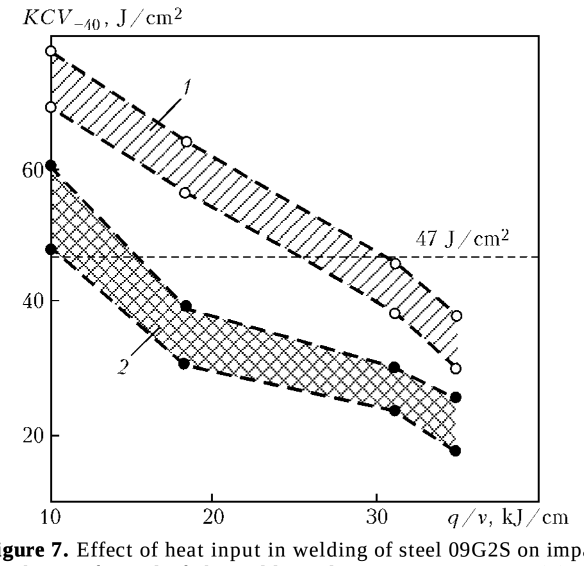


**Figure 7.** Effect of heat input in welding of steel 09G2S on impact toughness of metal of the welds made in Ar + 25 % $CO_2$ (*1*) and $CO_2$ (*2*)

And, thirdly, using welding wires with a decreased content of sulphur (≤ 0.01 wt.%) is a very efficient way of increasing cold resistance. For example, the content of sulphur in one of the wires Sv-08G2S used in our studies was 0.01 wt.%. Impact toughness of the weld metal produced by using this wire was much higher than in the case of wire with a sulphur content of about 0.02 wt.% (Figure 8).

At a low content of sulphur, the sulphide phase in the weld metal does not form continuous extended film precipitates [13] at the primary austenite grain boundaries, which might act as brittle fracture centres under dynamic loading. Furthermore, in formation of oxysulphide inclusions the sulphide phase does not form a continuous layer on the surface of oxide inclusions, i.e. it does not deactivate this surface that is the centre of epitaxial nucleation of acicular ferrite in $\gamma \rightarrow \alpha$ transformation occurring in cooling of the welds.

Therefore, in gas shielded welding of low-alloy steels, the available range of welding consumables (silicon-manganese wires and shielding gases) and set of technological approaches allow selection of such versions of the parameters that provide the required level of cold resistance of the weld metal, along with the optimal combination of other mechanical properties.

Replacement of one welding method by the other always requires optimisation of the welding technology and procedure, as well as evaluation of mechanical properties and performance of welded joints. Results of these investigations concerning welding of steels in argon mixtures are given in studies [7, 9, 10, 14].

Improvement of ductility, impact toughness and fracture resistance of the welds made in the Ar + $CO_2$ gas mixtures is attributable to the fact that the main structural component (70–75 vol.%) in this case is fine-grained acicular ferrite (Figure 9, *a*), in which microcracks develop in a zigzag manner because of the

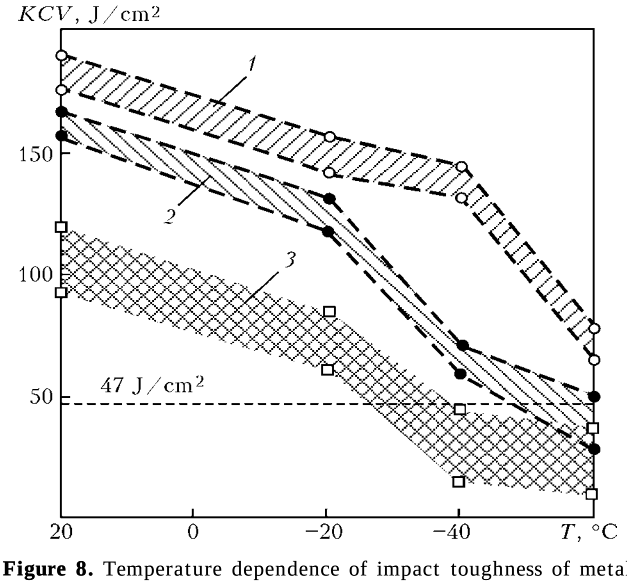


**Figure 8.** Temperature dependence of impact toughness of metal of the welds made on steel 09G2S using wire Sv-08G2S with different sulphur content: *1*, *2* — 0.011; *3* — 0.022 wt.% S; ○ — Ar + 25 % $CO_2$ welding; ●, ❐ — $CO_2$ welding

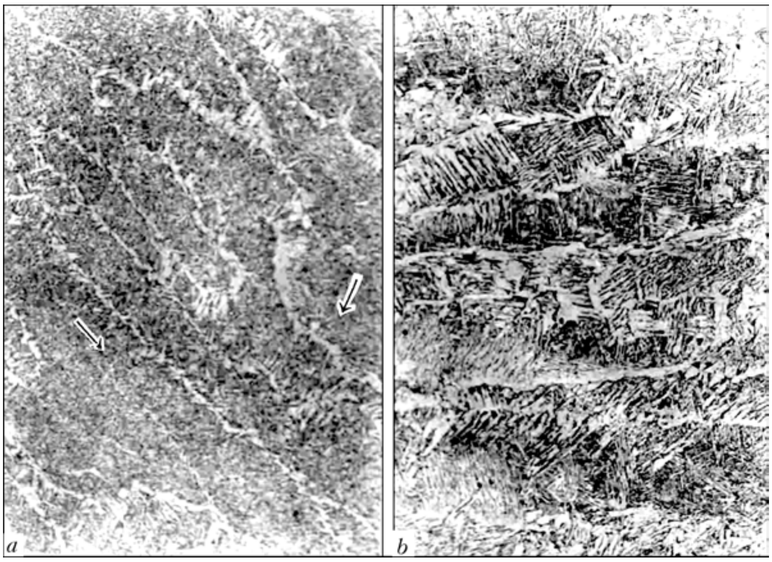


**Figure 9.** Microstructure of metal of the welds made in Ar + 25 % $CO_2$ mixture (*a*) and $CO_2$ (*b*); arrows show regions of acicular ferrite (×200)

presence of high-angle sub-grain boundaries [15]. Regions of hypoeutectoid ferrite and other structural components that embrittle metal at low temperatures are disconnected and form no continuous network along the primary austenite grain boundaries. The welds made in carbon dioxide (Figure 9, *b*) contain more hypoeutectoid ferrite and much less (30–40 vol.%) acicular ferrite.

Another important factor affecting toughness and ductility of the welds is the content and distribution of non-metallic inclusions. In the welds made in Ar + $CO_2$ mixtures, the content of oxygen and the associ-

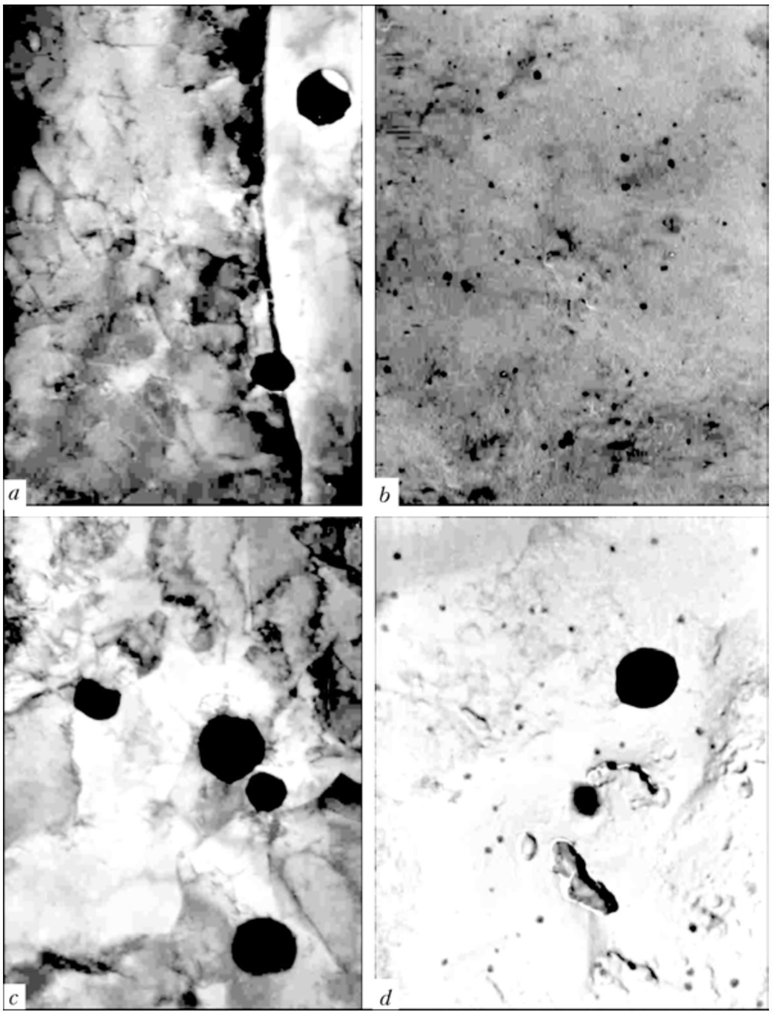


**Figure 10.** Fine structure of metal of the welds made in Ar + 25 % $CO_2$ mixture (*a*, *b*) and $CO_2$ (*c*, *d*), revealed by using fine foils (*a*, *c* — ×1350) and extraction carbon replicas (*b*, *d* — ×20000)

ated volume content of non-metallic inclusions are 2 times as low as in $CO_2$ welding.

In addition to inclusions detected by optical microscope, there are also sub-microscopic inclusions revealed in extraction carbon replicas by electron microscope [8]. These inclusions in metal of the welds made in the Ar + $CO_2$ mixture are uniformly distributed (Figure 10, *b*) and form no big clusters around coarse (1–2 μm) oxide inclusions, as is the case of the welds made in carbon dioxide (Figure 10, *d*). The absence of big clusters of fine inclusions near coarse oxide inclusions results in such a character of movement of dislocations in plastic deformation that leads to increase in fracture toughness of metal under conditions of positive and negative temperatures. This is proved by the data on the dislocation density measured on photos made by transmission electron microscope JEM-120 when examining fine foils. Metal of the welds made in the Ar + $CO_2$ gas mixture has the dislocation density equal to $3.4{\cdot}10^9$ cm$^{-2}$, and that made in carbon dioxide has the dislocation density equal to $5.0{\cdot}10^{10}$ cm$^{-2}$.

Moreover, multiple, uniformly distributed sub-microscopic oxide particles may act in cooling as ferrite centres in the austenite matrix [12]. As a result, $\gamma \to \alpha$ transformations create conditions for formation of a disoriented fine-grained structure of acicular ferrite, which is characterised by high ductility, cold resistance and resistance to brittle fracture.

Increase in brittle and tough fracture resistance of metal of the welds made in the Ar + $CO_2$ mixture is attributable to increase in the content of acicular ferrite, absence of coarse precipitates of the phases that embrittle metal at negative temperatures, and uniform distribution of non-metallic microscopic and sub-microscopic inclusions. Our data and results obtained by other researchers [16, 17] indicate that values of mechanical properties and crack resistance of metal of the welds made in argon-based gas mixtures correspond to the requirements imposed on joints and structures operating under the conditions of negative temperatures, dynamic loads and other unfavourable factors.

The Company «Azovmash» manufactured two pilot samples of bolsters using the developed technology. The bolsters were fatigue tested by the Dnepropetrovsk National University of Railway Transport using facilities of the Yuzhny Machine Building Factory. The load-carrying floor was used to secure the bolsters, and loading was performed by means of hydraulic cylinders. A macrocrack 30 mm long was detected in testing of the first bolster after 2,745,000 loading cycles, and loss of the load-carrying capacity occurred at 2,800,000 loading cycles. Initiation and development of a fatigue crack took place under the effect of normal stresses, which explains to a substantial degree the high rate of its propagation.

No fatigue macrocrack was detected in the second bolster after $6{\cdot}10^6$ loading cycles. Therefore, the tests were stopped at that. Their results provided a strong evidence of the possibility of making welded bolsters.

## CONCLUSIONS

1. The developed welded freight bogie bolster meets all the requirements imposed on welded structures operating under cyclic loading.

2. The recommended technology for manufacture of bolsters is based on performing welding in the Ar + 25 % $CO_2$ mixture. Bolsters made by this technology meet all the requirements imposed on structures operating under conditions of negative temperatures, dynamic loads and other unfavourable factors.

3. As shown by the tests of pilot samples of the welded bolsters, they have the required fatigue life, are not inferior to the cast ones in strength and technical-economic indicators, and can be recommended for comprehensive cyclic tests.


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# ROLE OF PEAK STRESSES IN FORMATION OF COLD CRACKS IN WELDED JOINTS OF HARDENABLE STEELS

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The paper presents analysis of the most important factors promoting cold cracking in welded joints of hardenable steels. It is shown that the most important components causing cold cracking are «peak» stresses in as-hardened metal structure and level of hydrogen content in points of local «peak» stresses.

*Keywords: peak stresses, relaxation, microplastic deformation, hydrogen, delayed fracture, cold cracks, residual stresses, variable stresses*

Fabrication of welded structures from high-strength hardenable steels involves certain difficulties because of the brittle fracture susceptibility of welded joints. This is attributable to the high probability of formation of such defects as cold cracks in the metal of welded joint HAZ.

Many researchers note in their work three main factors, essentially influencing the process of cold cracking, namely:

- structural state of welded joint metal characterized by the presence of martensite and bainite type components;
- concentration of diffusible hydrogen in the zone of crack site initiation;
- level of stresses in the welded joint.

On the other hand, the opinions of the researchers on crack nuclei formation and their location are significantly different. M.Kh. Shorshorov believes that formation of delayed fracture cracking by Zener schematic does not provide a satisfactory explanation for this process [1]. Zener schematic used by S.S. Shurakov [2] and A.M. Makara [3], explains development of significant tensile stresses and formation of crack nuclei in grain regions adjacent to boundary junction as a result of plastic flow (boundary shift). However, M.Kh. Shorshorov [1] assumes that vacancy sink to stress concentration zones has an essential role in the mechanism of crack nuclei formation along the grain boundaries. E.L. Makarov [4] believes that cold cracking starts with formation of a fracture site on the boundaries of austenite grains in the HAZ metal section adjacent to the fusion zone.

Many researchers still are of the opinion that the lower the level of residual stresses in the welded joints of hardenable steels, the lower is the probability of cold crack formation in them. It should be noted, however, that in a number of cases cold cracking is observed also in welded joints made under the conditions providing minimum residual stresses, the level of which is essentially lower than that of the critical stresses [5]. It is known that a number of studies [6, 7] report the great influence of the so-called peak stresses on the cold cracking processes in steel that was not subjected to further heat treatment. In [6] it is shown that local peak stresses, greatly superior to average stresses, may develop in as-hardened structure of the metal with the start of plastic deformation [8, 9], that may result in initiation and propagation of fracture cracks. It is also established that at sample loading below the macroscopic yield point microplastic deformation develops in the regions of the action of peak stresses, leading to relaxation of these stresses and increasing the brittle fracture resistance.

In [7, 9] it is noted that peak stresses develop in microvolumes in hardened steel as a result of the shear nature of martensite transformation, their level greatly exceeding the average level of residual stresses in the regions of localization. Local peak stresses are mainly induced in the points of the martensite crystal meeting the boundary of the initial austenitic grain, as a result of dynamic impact of the rapidly growing martensite crystal on this boundary [10]. They may be assumed to be the main cause for delayed fracture, as fracture of hardened steels proceeds under the impact of static loading, the level of which is below the yield point.

Formation of local peak stresses causing crack initiation, is explained in the description of mechanisms of dislocation models developed by Kottrell and authors, for instance, Straw–Mott [8]. According to the latter, flat clusters of sliding dislocations form in front of various obstacles, resulting in a strong local stress concentration, sufficient for cracking.

Thus, peak stresses in as-hardened steel are one of the main causes for formation of local cold crack nuclei. Thermal cycle of welding and welding consumable composition are extremely important in ensuring the welded joint resistance to cold cracking. In the case of application of welding consumables close to the base metal by the level of alloying, welding of hardening steels is performed with preheating and subsequent heat treatment to alleviate the adverse effect of the above factors. However, the above technological operations are labour-consuming and in a number of cases inacceptable (for instance, in welding of large-sized casing structures). Therefore, high-alloyed welding consumables, providing an austenitic weld metal, are widely applied now for welding of the above steels

[11], this allowing an essential improvement of crack resistance, without, however, ensuring the equivalent strength of welded joints. Equivalent strength of the welded joint is achieved by applying welding consumables of a new generation, providing high-alloyed martensitic or austenitic-martensitic welds, which by their strength characteristics are not inferior to the base metal in welding without preheating or heat treatment [12].

**Influence of weld metal type on the structure and properties of the HAZ metal.** There exist various interpretations of the influence of weld metal alloying, protective coating and shielding gas composition on the resistance of the welded joint HAZ metal to cold cracking.

When welding consumables are used, which provide ferrite welds, overheating of the HAZ metal and its staying in the temperature region of intensive homogenizing of austenite for a rather long time are observed [13, 14]. As a consequence, the temperature of the start of overcooled austenite transformation in the HAZ metal is essentially reduced ($T_{Ms}$ < 250 °C), this promoting formation of brittle hardening structures (highly tetragonal martensite, lower bainite, as well as their mixtures). The above negative phenomena are eliminated, as a rule, by preheating and subsequent heat treatment of the welded joint. If a high-alloyed austenitic weld metal is produced, the level of HAZ metal overheating is reduced due to its lower solidification temperature, as a result of a shorter time of staying at a high temperature. This circumstance promotes a higher inhomogeneity of austenite and, thus causes an increase of temperature of the start of its transformation.

In [15, 16] it is noted that during cooling of the austenite weld metal plastic deformation grows continuously right up to temperatures, corresponding to the start of martensite transformation in the HAZ metal. The latter causes a more active transformation of austenite in the HAZ metal and shifting of temperature of the bulk of $\gamma \rightarrow \alpha$ transformation to the region of higher temperatures, this promoting development of structures resistant to cold cracking in the HAZ (tempering martensite, upper bainite, as well as their mixtures).

High-alloyed low-carbon martensitic and austenitic-martensitic welds behave similar to austenitic ones at cooling (while having all of their advantages), i.e. preserve an austenitic structure at cooling to temperatures, lower compared to the temperature interval of structural transformations in the HAZ metal ($T_{Ms}$ < < 200 °C), this, similar to the previous case, promoting the shifting of the main bulk of structural transformations in the HAZ metal to the high temperature region. As a result of subsequent $\gamma \rightarrow \alpha_M$ transformation the weld acquires a structure, the main component of which is high-alloyed low-carbon martensite, giving the weld a high strength ($\sigma_t \geq 1000$ MPa). In addition, the above transformations proceed with increase of volume, this promoting partial relaxation of residual stresses in the welded joint.

The above reasoning is based on the data obtained by different methods of studying and procedures of testing welded joints on hardenable steels. Tekken, Lehigh University, CTS, cross, RD samples and procedure with additional external load, Implant sample, etc., were used as the latter.

The disadvantage of the above procedures is the fact that they mainly reply to the questions of whether there are cracks, or not, assess their extent, and some procedures also determine critical stress $\sigma_{crit}$, at which the welded joints do not fail for 20 h of their testing. The above disadvantages of the currently available procedures of evaluation of the technological strength of welded joints account for the rationality of developing new and further improvement of the existing procedures of studying the kinetics of formation of the structure and stressed state of the welded joint, that would provide new information on the factors and the nature of their influence on the processes, proceeding at cold crack formation.

**New procedure of investigation and testing of welded joints.** Compared to the existing methods, the new procedure has several differences [17] and allows:

- evaluation of the resistance of welded joint HAZ metal to cold cracking using a new quantitative criterion — energy consumption of the fracture process;
- studying the kinetics of phase transformation in weld metal and HAZ and nature of the change of the stressed-strained state during cooling of the welded joint;
- obtaining data on plastic deformation in the fusion zone during cooling of the welded joint;
- studying the influence of weld metal on the magnitude of time and residual stresses;
- evaluation of the nature of welded joint rigidity (with different processes of welded sample fastening — in the unrestrained condition or at rigid restraint in the loading device) on its delayed fracture resistance;
- assessment of the influence of additional load, applied at different temperatures during welded joint cooling, on the delayed fracture resistance of the tested samples.

**Investigation of cold cracking resistance of welded joints on hardenable steels.** Investigations performed using the new procedure, allowed not only evaluation of the results of earlier work [13, 14, 18], but also obtaining new data on the features of welding of hardening steels by the ferritic-pearlitic, austenitic and austenitic-martensitic welding consumables. As was noted earlier, austenitic welding consumables have become accepted in many branches of the national economy to increase the cold cracking resistance of welded joints on hardenable steels.

The purpose of these studies is evaluation of the role of the welded joint stressed-strained state, taking into account peak stresses as one of the main factors, affecting the mechanism of initiation and propagation of cold cracks in the metal of the HAZ of welded joints on hardenable steels.

**Table 1.** Mechanical properties of the studied steels

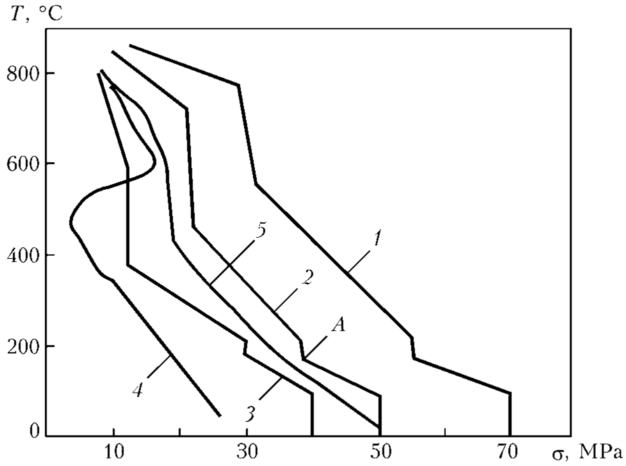
| Steel grade | $\sigma_{0.2}$, MPa | $\sigma_t$, MPa | $\delta_5$, % | $\psi$, % |
|---|---|---|---|---|
| 12Kh2N3M | 600 | 800 | 20 | 60 |
| 14KhGN2MDAFB | 750 | 860 | 20 | 65 |
| 30KhGSA | 1000 | 1100 | 15 | 25 |
| 15Kh2N4MDA | 820 | 860 | 20 | 60 |

The main attention in experimental investigations [17, 19] was given to the important role of the time and residual stresses, developing during welded joint cooling. It is established that the time and residual stresses of a certain level have a pronounced positive influence on the cold cracking resistance of HAZ metal of welded joints on alloyed steels.

Further investigations were conducted on samples of hardenable steels 12Kh2N3M, 15Kh2N4MDA, 14KhGN2MDAFB, 30Kh2N2M with inserts from steels 30KhGSA, 12Kh2N3M, 15Kh2N4MDA, 14KhGN2MDAFB and 30Kh2N2M. Mechanical properties of the studied steels are given in Table 1. Welding was performed with welding wires Sv-10Kh2GSMA (ferritic-pearlitic), Sv-Kh20N9TG7T (austenitic) and test low-carbon austenitic-martensitic flux-cored wires PP-ANVP-60 (05Kh13N9M2GST), PP-ANVP-80 (05Kh12N8M2GST) and PP-ANVP-100 (05Kh10N7GST). Designations of test wires are given in keeping with the weld metal yield point (600, 800 and 1000 MPa). Welding of test samples was performed by mechanized process in $CO_2$ in the following mode: welding current of 280 to 300 A, arc voltage of 30 to 32 V, welding speed of 8, 12 and 16 m / h, this corresponding to HAZ metal cooling rate of 6, 7, 10 and 14.5 °C / s in the temperature range of 600–400 °C.

Results of testing the studied samples for delayed fracture resistance are given in Tables 2–8. Table 2 gives the results of testing samples from steel 12Kh2N3M with inserts of the same steel. Welding was performed in the same mode with austenitic-martensitic electrode wires. Values of residual stresses were equal to 40–70 MPa. The Figure gives the graphs of phase transformations in welding with the above wires (curves *1*–*3*). Samples after testing at the load of $P$ = 650 MPa did not break up during 20 h.

To compare the delayed fracture resistance of welded joints, samples from steel 12Kh2N3M were welded with inserts from steel 30Kh2N2M by ferritic-




Kinetics of phase transformations in the weld metal depending on wire type: *1* — PP-ANVP-60; *2* — PP-ANVP-80; *3* — PP-ANVP-100; *4* — Sv-10KhG2SMA; *5* — Sv-08Kh20N9G7T

pearlitic Sv-10KhG2SMA, austenitic Sv-Kh20N9G7T and austenitic-martensitic PP-ANVP-80 wires. Welding was conducted in the same modes and at the same load. Time to fracture of tested samples and residual stress values are given in Table 3.

In [13, 18, 19] it is shown that increase of the delayed fracture resistance of HAZ metal of welded joints can be achieved using additional load during cooling of the tested samples. This process was used also in this study. Investigation results are given in Tables 4 and 5. As is seen from Table 4, items 1 and 5, in welding of unrestrained samples their resistance is quite low in the absence of the time and residual stresses. Time-to-fracture is somewhat increased in the case of 50 MPa load application before testing of such samples for delayed fracture (items 3 and 7). At testing of restrained samples, a sample did not fail at the additional load of 120 MPa during cooling at the temperature of 300 °C (items 4 and 8).

Table 5 gives the results of delayed fracture testing of samples from steel 14KhGN2DAFB, in welding of which the HAZ metal formed a structure with a large amount of martensite (cooling rate of 14.5 °C / s). Values of residual stresses were equal to 95 and 106 MPa (items 1 and 2), this being insufficient to prevent sample fracture during 20 h. The Table also gives the results of delayed fracture testing of samples at their additional loading after their cooling to 50 °C, or to room temperature. At additional load of up to 200 MPa and soaking for 30 min before the start of testing for delayed fracture at the load of 600 MPa a sample (item 3) failed after 12 h 5 min, at soaking for 3 h — after 14 h 45 min (item 4). At additional

**Table 2.** Results of investigations of delayed cracking resistance of welded joints ($U_a$ = 32 V, $I_w$ = 300 À, $v_w$ = 16 m / h, Đ = = 450 MPa)

| No. | Sample material | Insert material | Flux-cored wire grade | Residual stresses, MPa | Time to fracture, min |
|---|---|---|---|---|---|
| 1 | 12Kh2N3M | 12Kh2N3M | PP-ANVP-60 (05Kh13N9M2G2ST) | 70 | No fracture |
| 2 | | | PP-ANVP-80 (05Kh12N8M2GST) | 50 | Same |
| 3 | | | PP-ANVP-100 (05Kh10N7GST) | 40 | » |

**Table 3.** Influence of electrode wire composition on delayed fracture resistance of welded joints ($U_a$ = 32 V, $I_w$ = 300 À, $v_w$ = 16 m / h, *Đ* = 450 MPa)

| No. | Welding wire grade | Residual stresses, MPa | Time to fracture, min |
|---|---|---|---|
| 1 | Sv-10KhG2SMA | 45 | 15 |
| 2 | Sv-Kh20N9G7T | 50 | No fracture |
| 3 | Sv-05Kh12N8M2GST | 50 | Same |

loading of a sample up to 300 MPa, and soaking for 3 h the sample did not fail (item 5), so that the experiments on improvement of sample resistance by selection of the respective welding mode were continued (Table 6).

As a result of the change of welding mode, the cooling rate dropped from 14.5 to 6 °C / s. Value of residual stresses was increased, HAZ metal structure was improved, amount of martensite was reduced, hardness was decreased and the samples did not fail. Table 7 gives the results of studying the influence of application of austenitic wire Sv-08Kh20N9G7T on delayed fracture resistance of samples from steel 15Kh2N4MDA with 30KhGSA insert. In welding in the mode ensuring the cooling rate of 6 °C / s, the samples did not fail in those cases, when the value of residual stresses was equal to 115–120 MPa. At 14.5 °C / s rate of HAZ metal cooling and residual stresses not higher than 65 MPa, time-to-fracture of such samples was from 5 to 31 min.

Tables 5 and 8 show that the time of soaking or recovery of tested samples after welding before their loading to specified load has a strong influence on delayed fracture resistance of welded joints. As a result of additional loading of samples after welding to 200 MPa and soaking for 1 h, time-to-fracture was equal to 151 min at the load of 650 MPa (Table 8, item 2), and after soaking for 48 h the sample did not fail at the same load. In welding of samples, the cool-

**Table 4.** Influence of additional load on delayed fracture resistance of welded joints ($U_a$ = 32 V, $I_w$ = 300 À, $v_w$ = 16 m / h, *Đ* = = 450 MPa)

| No. | Sample material | Insert material | Wire grade (type) | Temperature of application and value of additional load | Residual stresses, MPa | Time to fracture, min |
|---|---|---|---|---|---|---|
| 1 | 12Kh2N3M | 30Kh2N2M | PP-ANVP-80 (05Kh12N8M2GST) | – | Sample is not fastened | 30 |
| 2 | | | PP-ANVP-80 (05Kh12N8M2GST) | – | 75 | No fracture |
| 3 | | | PP-ANVP-80 (05Kh12N8M2GST) | At 50 °C up to 50 MPa | 50 | 2 h |
| 4 | | | PP-ANVP-80 (05Kh12N8M2GST) | At 300 °C up to 120 MPa | 120 | No fracture |
| 5 | | | EI-613 (Sv-08Kh20N9G7T) | – | Sample is not fastened | 35 |
| 6 | | | EI-613 (Sv-08Kh20N9G7T) | – | 130 | No fracture |
| 7 | | | EI-613 (Sv-08Kh20N9G7T) | At 50 °C up to 50 MPa | 50 | 14 h 35 min |
| 8 | | | EI-613 (Sv-08Kh20N9G7T) | At 300 °C up to 120 MPa | 120 | No fracture |

**Table 5.** Influence of additional load on delayed fracture resistance of welded joints made with PP-ANVP-80 wire ($U_a$ = 32 V, $I_w$ = = 300 À, $v_w$ = 16 m / h, *Đ* = 600 MPa)

| No. | Sample material | Insert material | Temperature of application and value of additional load | Soaking time | Residual stresses, MPa | Time to fracture |
|---|---|---|---|---|---|---|
| 1 | 14KhGN2DAFB | 14KhGN2DAFB | – | – | 95 | 5 h 25 min |
| 2 | | | – | – | 106 | 15 h 25 min |
| 3 | | | At 50 °C up to 200 MPa | 30 min | 200 | 12 h 5 min |
| 4 | | | At 50 °C up to 200 MPa | 3 h | 200 | 14 h 45 min |
| 5 | | | At 50 °C up to 300 MPa | 3 h | 300 | No fracture |

**Table 6.** Influence of welding mode on delayed fracture resistance of welded joints made with PP-ANVP-80 wire ($U_a$ = 32 V, $I_w$ = = 300 À, m / h, *Đ* = 600 MPa)

| No. | Sample material | Insert material | Welding speed, m / h | Residual stresses, MPa | Time to fracture |
|---|---|---|---|---|---|
| 1 | 14KhGN2DAFB | 14KhGN2DAFB | 8 | 180 | No fracture |
| 2 | 14KhGN2DAFB | 14KhGN2DAFB | 12 | 140 | 15 h 25 min |
| 3 | 14KhGN2DAFB | 14KhGN2DAFB | 16 | 120 | 5 h 25 min |

**Table 7.** Influence of residual stresses on delayed fracture resistance of welded joints ($U_a$ = 32 V, $I_w$ = 270 À)

| No. | Sample material | Insert material | Wire type | Welding speed, m / h | Test load, MPa | Length of sample-insert, mm | Residual stresses, MPa | Time to fracture, min |
|---|---|---|---|---|---|---|---|---|
| 1 | 15Kh2N4MDA | 30KhGSA | Sv-08Kh20N9G7T | 8 | 450 | 140 | 95 | 540 |
| 2 | | | | 8 | 450 | 90 | 115 | No fracture |
| 3 | | | | 8 | 550 | 70 | 120 | Same |
| 4 | | | | 16 | 550 | 140 | 30 | 5 |
| 5 | | | | 16 | 550 | 85 | 60 | 25 |
| 6 | | | | 16 | 550 | 70 | 65 | 31 |

ing rate of which was 14.5 °C / s, and time of soaking before loading was 12 min, fracture occurred after 110 min at the load of 550 MPa, and at soaking for 120 h after welding the sample did not fail at the same load.

**Analysis of investigation results and evaluation of the role of peak stresses.** Analysis of the results of studying the energy consumption of the cold cracking process in welded joints of hardenable steels is indicative of the important role of the time and residual stresses in their resistance to delayed cracking. It is established that at residual stresses on the level of 80–160 MPa (stresses depend on the strength characteristics and structure of steel), resulting from welding of hardenable steels or additional loading of samples during cooling or after their complete cooling, the delayed fracture resistance of welded joints is essentially increased, this being confirmed by a considerable increase of the required work consumed in fracture.

The above leads to the assumption that peak stresses which are much higher than the average level of stresses in as-hardened HAZ metal, are observed in the HAZ metal of hardenable steel welded joints. Development of high local microstresses is one of the main causes for formation of crack nuclei in the HAZ metal. It is known that peak stresses may relax as a result of metal heating, or application of additional load during or after cooling of the welded joint or as a result of metal recovery. Studies showed that to prevent cold crack formation in the HAZ metal of welded joints, peak stresses have to be relaxed during or right after welding before design load application. Peak stress relaxation can be implemented as a result of preheating of the edges being welded, or local or total heat treatment of the welded joint. Edges preheating in welding, application of welding modes providing a lower cooling rate, promote improvement of the structure of weld and HAZ metal, increase of shrinkage stresses, more complete removal of hydrogen from the welded joints. Such measures allow increasing the welded joint resistance to cold cracks. Heat treatment in furnaces or local heat treatment is performed for tempering of welded joints after welding. Such a treatment promotes relaxation of peak stresses, improvement of the structure and increase of ductile properties and toughness of welded joint metal. For relaxation of peak stresses it is also possible to apply the developed process of welding with heat treatment [20], involving welding of hardenable steels with welding wires providing welds of a ferritic-pearlitic structure and re-heating of weld metal by a non-consumable electrode in inert gas. Re-heating leads to a significant reduction of weight fraction of hydrogen in the weld metal, and lowering of its cooling rate, which results in improvement of HAZ metal structure, and increase of the rate of peak stress relaxation. Developed method allows controlling the welded joint cooling rate in a broad range. Re-heating and partial melting of weld metal, deposited with a consumable electrode, promote formation of residual stresses sufficient for relaxation of peak stresses and significant lowering of hydrogen concentration in the weld and HAZ metal. This method provides the required quality of hardenable steel welded joints, using low-alloyed welding consumables for welding. Its application allows reducing the labour consumption and cost of welding operations, compared to the processes of welded joint heat treatment applied now. Welding

**Table 8.** Influence of soaking (recovery) of welded joints made with Sv-10KhG2SMA wire on the delayed fracture resistance ($U_a$ = = 32 V, $I_w$ = 300 À)

| No. | Sample material | Insert material | Welding speed, m / h | Additional load up to the value of, MPa | Soaking time, h | Testing load, MPa | Time to fracture, min |
|---|---|---|---|---|---|---|---|
| 1 | 12Kh2NMFA | 30KhGSA | 8 | – | – | 650 | 40 |
| 2 | 14KhN3MDA | | 8 | 200 | 1 | 650 | 151 |
| 3 | 12Kh2NMFA | | 8 | 200 | 48 | 650 | No fracture |
| 4 | 14KhN3MDA | | 16 | – | – | 550 | 70 |
| 5 | 14KhN3MDA | | 16 | 200 | 12 min | 550 | 110 |
| 6 | 14KhN3MDA | | 16 | 350 | 120 | 550 | No fracture |

consumables of austenitic or austenitic-martensitic type can also be used for this purpose. Application of such consumables in many cases allows eliminating expensive and labour-consuming operations associated with heat treatment. In addition, application of austenitic-martensitic welding consumables allows producing welded joints, where the weld metal mechanical properties are close to those of the base metal. Use of the above consumables for welding is characterized by the required level of time and residual stresses and a more favourable influence of the austenitic or high-alloyed low-carbon martensitic weld metal on the HAZ metal structure.

**Hydrogen influence on cracking.** Alongside peak stresses, hydrogen content in welded joint metal also has a great influence on crack initiation and propagation.

In [21] it is noted that in the opinion of many researchers, hydrogen has the ability to diffuse into the zone of tensile stress concentration, i.e. into locations, where steel is in the condition of uniform tension. At concentration of a critical amount of hydrogen in them a crack is formed. In [22, 23] it is stated that in the zone of tensile stress concentration and microscopic deformation hydrogen content increases as a result of its interaction with a dislocation cluster at the concentrator tip. Hydrogen penetration into this zone as a result of its drift diffusion may cause hydrogen concentration, which is several times higher than its average content in the metal [24]. Hydrogen accumulation can cause cracking. Analysis of the features of development of a local non-uniformity of hydrogen distribution in the welded joint by the moment of crack initiation has a decisive importance for studying the conditions of cold cracking (delayed fracture), alongside with allowing for microlocal plastic deformation and critical level of bulk tensile stresses [23].

In [25] a new physical model of hydrogen embrittlement has been proposed, in keeping with which «in the absence of irreversible traps (pores, cracks, interphases boundaries) hydrogen is condensed in the traps, which are the dislocations. At plastic deformation hydrogen is transported by moving dislocations to the crack initiation site. Microcrack initiation can be presented by the classic Zener–Straw model: a microcrack forms in the tip of the cluster of dislocations arrested by the grain boundary or other obstacles».

In [26] it is noted that hydrogen and the associated delayed fracture of welded joints is the most clearly manifested in arc welding of alloyed steels with limited (to 0.2 %) carbon content. This is due to the fact that hydrogen is a unique technological impurity with a high mobility in the crystalline lattice and the capability of segregation on the boundaries as a result of transportation by moving dislocations at temperatures below 400 °C. The extent of development of the delayed fracture process depends chiefly on the thermodeformational welding cycle, phase transformations of the base and weld metal, condition of grain boundaries and residual stresses [27].

Increased content of hydrogen in the high-strength high-alloyed weld metal has a negative influence on its mechanical properties (primarily, on ductility), and in a number of cases promotes cold cracking in the welded joints [5]. When studying hydrogen influence on the mechanical properties of multilayer martensitic weld metal of 08Kh12N8M2GST type in welded joints of 15Kh2N4MDA steel, which were made by coated-electrode manual arc welding, it was established that at increase of the amount of diffusible hydrogen from 5.3 to 9.0 $cm^2$ per 100 g of the metal, the level of ductile properties of the high-strength weld metal decreases significantly ($\delta$ = 7–9 %, $\psi$ = = 13–14 %). Transverse and longitudinal cold cracks can form in the weld metal. At diffusible hydrogen content of 3.2–3.5 $cm^2$ per 100 g of metal no cold cracks were found in the welded joint. Evaluation of the technological strength of welded joints by welding up rigid TsNIITS samples showed the high resistance of welded joints to cold cracking at a total content of hydrogen in the weld metal of not more than 3.5 $cm^3$ per 100 g of metal. In the case of weld metal saturation by hydrogen up to 5.3 $cm^3$ per 100 g of metal, transverse (in the fusion zone) and longitudinal (in the center) cold cracks are found in the welds. Lower moisture content in the electrode coating due to correction of its composition and use of optimum baking temperature provided a low content of diffusible hydrogen in the weld metal of 2.3 $cm^3$ per 100 g of metal. It is established that the weld metal ductile properties and its cold cracking resistance were essentially improved.

Studied was the influence of the amount of hydrogen in the weld, depending on baking temperature of electrodes of austenitic-martensitic type, on the delayed fracture resistance of the HAZ metal of welded joints of steel 12Kh2N3M. Table 9 gives the experimental data. In view of a strict limit on the amount of hydrogen in austenitic-martensitic welds for welded joints of high-strength hardenable steels, the E.O. Paton Electric Welding Institute developed a group of welding consumables, namely electrodes of ANVP-80 type for manual arc welding and flux-cored wire of PP-ANVP-80 type for mechanized $CO_2$ welding. The above welding consumables ensure hydrogen concentration in welds on the level of not more than 3.0 $cm^3$ per 100 g of the metal [5].

**Measures promoting an increase of delayed fracture resistance of welded joints.** Investigations showed that to prevent cold cracking and reduce the susceptibility of as-hardened structure of HAZ metal to their formation, it is necessary to take measures for relaxation of peak stresses even before the moment of applying the working load to the tested sample. This means that residual stresses in the welded joint required for relaxation of peak stresses in the HAZ metal should be not lower than a certain limit. Such a relaxation proceeds by microplastic shear (at a sufficient mobility of dislocations and increased loading temperature). At further increase of the residual stresses

**Table 9.** Hydrogen influence on delayed fracture resistance of welded joints made with ANVP-80 electrodes

| No. | Temperature of electrode baking for 2 h, °C | Hydrogen concentration in weld metal, $cm^3$ per 100 g of metal | | Time to fracture | Remark |
|---|---|---|---|---|---|
| | | Calculated | Actual | | |
| 1 | 350 | 5.5 | 5.7 | 116 min | – |
| 2 | 350 | 5.5 | 5.4 | 118 min | Soaked at 185 MPa for 2 h at 300 °C |
| 3 | 450 | 4.5 | 4.6 | 11 h 15 min | – |
| 4 | 450 | 4.5 | 4.5 | 18 h 40 min | Soaked at 185 MPa for 2 h at 300 °C |
| 5 | 500 | 3.0 | 3.4 | 28 h | – |
| 6 | 550 | 2.5 | 2.8 | No fracture | – |

at a limited mobility of dislocations or lower loading temperature, another method of relaxation is implemented, leading to initiation of microcracks increasing the risk of brittle fracture [7, 9].

To determine the optimum value of residual stresses, required for relaxation of peak stresses through microplastic deformation, experimental studies should be done by the new procedure, as their value depends on the composition of steel, used for the welded structure, welding consumables and welding technology.

Experimental data are indicative of the favourable influence of austenitic or austenitic-martensitic wires and electrodes on the delayed fracture resistance of the HAZ metal of welded joints of hardenable steels. Welding with such consumables promotes increase of the values of residual stresses in the studied samples by approximately 2 times, compared to the values of residual stresses characteristic for welding with ferritic-pearlitic wires (see Figure, curves *2*, *4*). It is important to note that the values of displacement (shrinkage) and residual stresses in welding with austenitic (Sv-Kh20N9G7T) and austenitic-martensitic wire (Sv-05Kh12N8M2GST) are quite close (see Figure, curves *2*, *5*). This is attributable to the fact that the weld metal deposited by austenitic-martensitic wires, has an austenitic structure before the start of martensitic transformation at $T$ = 140--160 °C (point $A$). At further cooling almost to room temperature (point $B$) the weld metal contains a still quite large amount of austenite.

In welding of hardenable steels, depending on the used welding modes (thermal cycles) a bainite, bainite-martensite and martensite structure may form in the HAZ metal. For instance, at the welding speed $v_w$ = 16 m / h the welded joint HAZ metal may form a structure with a higher martensite content compared to the HAZ metal of a welded joint made at a lower welding speed ($v_w$ = 8 m / h), this being attributable to a higher cooling rate. The HAZ metal forms a structure with a large amount of martensite, which is confirmed by a high level of microhardness ($HV$ 2700–2900 MPa). For relaxation of peak stresses such a structure requires a higher level of residual stresses, as it is difficult for microplastic deformation to proceed here. In this case it is necessary to use welding modes, providing a lower cooling rate. Such a measure promotes, first of all, increase of the extent of shrinkage of the welded joint and increase of residual stresses to the specified level, and, secondly, formation of a more favourable structure of upper bainite or a mixture of upper bainite with tempering martensite in the HAZ metal. In welds of austenite-martensite type the volume effect at $\gamma \rightarrow \alpha_M$ transformation is less pronounced. Martensite transformation in the HAZ metal in the case of a ferritic weld deposition is accompanied by a much greater volume effect, which may lead to a greater deformation on the fusion boundary of the formed martensite with the HAZ metal structure and cause a high level of microstresses. Investigation results confirm the above assumptions. Additional experiments are required for a numerical substantiation.

Investigation results described in this paper, have a not only theoretical, but also practical importance, as the found tendencies are characteristic also for welded joints of the actual structures. For instance, technology envisaging preliminary compression after welding for relaxation of peak stresses even before their loading up to the specified working load can be recommended to increase the cold cracking resistance of welded structures (pipelines, vessels, tanks, cylinders, etc., particularly thin-walled ones from hardenable steels) during their fabrication. Compression pressure should be given in the specification for this structure fabrication. Magnitude of stresses caused by compression should be determined, depending on materials used for the structure fabrication, its purpose and fabrication technology.

In conclusion it should be noted that the main cause for nucleus crack formation are peak stresses, resulting from local clustering of dislocations and others defects in as-hardened structure of the HAZ metal, which is confirmed by a number of studies of the known authors. Increase of hydrogen concentration in the HAZ metal and its accumulation in the sites of the local peak stresses, promotes a lowering of the metal resistance to formation of crack nuclei and their further propagation. Relaxation of peak stresses leads to lowering of weight fraction of hydrogen in these local areas as a result of its diffusion to other regions of the HAZ metal (with a higher level of stress concentration). It may be assumed that relaxation of peak stresses will promote a lowering of hydrogen solubility in the local regions of the structure.

Formation of peak stresses promotes an interaction of the HAZ metal structure (which was not subjected to further heat treatment) and weld metal structure formed as a lower temperature. The higher the volume effect of the bainite-martensite transformation of the weld metal austenite, the stronger is the influence of such a structure on the HAZ metal. The source of such interaction forces creates strong microdeformations and higher local stresses [6].

Relaxation of peak stresses proceeds during welded joint cooling under the influence of the time and residual stresses, caused by metal shrinkage and phase transformations of the weld and HAZ metal. Relaxation of peak stresses in a welded joint during and / or after cooling is required before its loading by a specified load in delayed fracture testing of samples. Completeness of peak stress relaxation depends on residual stresses in a welded joint.

In case of formation of a bainite-martensite or martensite structure in the HAZ metal, relaxation of peak stress requires a higher level of residual stresses, as microplastic deformation is difficult in such a structure. The following currently applied methods can be used to reduce peak stresses:

• preheating of the edges of the items being welded; their heat treatment in furnaces or local heat treatment; application of welding modes, providing a lower cooling rate to increase the shrinkage stresses and form a favourable structure in the HAZ metal;

• welding of structures with austenitic and austenitic-martensitic welding consumables, which, compared to the ferritic ones, provide a weld metal characterized by a lower volume effect as a result of phase transformations and greater value of residual stresses in the welded joint. In addition, it should be noted that use of austenitic-martensitic welding consumables allows producing high-strength ($\sigma_t \geq$ $\geq$ 1000 MPa) welded joints of hardenable steels in welding without preheating and heat treatment;

• welding with additional arc heat treatment [20], which allows lowering the labour consumption and cost of welding operations compared to welding technologies, envisaging the traditional techniques of welded joint heat treatment.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# ENERGY CHARACTERISTICS OF LOW-AMPERAGE ARCS

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Energy characteristics of low-amperage arcs were studied. It was established that the low-amperage arcs in argon-arc welding are characterised by a higher (3–5 times) concentration of heat flow, compared with arcs in plasma-arc cladding. Heat input and heat flow concentration are the main parameters allowing a wide-range control of the weld pool formation and strained state of the HAZ metal.

*Keywords: argon-arc welding, plasma-arc cladding, effective thermal power, effective efficiency of heating of workpiece, heat flow distribution, heat input*

Welding methods providing a limited heat input into the base metal have been finding lately an increasing application for joining hard-to-join heat-resistant materials, which is the case, in particular, of repair [1–3]. In this connection, the possibility of using low-amperage arcs characterised by a minimal heat input into a workpiece, which raises operational strength of a material welded, is gaining in importance [4].

In view of a high cost of nickel- and cobalt-base materials, as well as difficulties associated with their machining, to reduce sensitivity to hot cracking it is expedient to use mathematical modelling of the weld formation to provide an efficient control of the process and optimise welding and cladding conditions. Size of the weld pool and intensity of phase, structure and volume transformations occurring in welding and cladding are determined, first of all, by the peculiarity of distribution of heat flow of the arc over the heating spot. The effective thermal power of the arc and radial distribution of the arc heat being known, it is possible to get an idea of the distribution of heat over the metal surface and size of the heating spot [5, 6].

Distribution of the specific heat flow on the radius of the heating spot can be described with a sufficient degree of accuracy by the Gauss law of normal probability distribution [7, 8] (Figure 1):

$$q_r = q_{\max} e^{-kr^2}; \qquad (1)$$

$$q_{\max} = \frac{q_i k}{\pi}, \qquad (2)$$

where $q_r$ is the surface density of the heat flow at distance $r$ from the heating spot centre, W / cm$^2$; $k$ is the concentration factor that depends upon the size and distribution of heat across the section of the heating spot, cm$^{-2}$ (terminology by N.N. Rykalin [7]); $q_{\max}$ is the surface density of the heat flow at the heating spot centre, W / cm$^2$; and $q_i$ is the effective thermal power of heating of an item, W.

When using expression (1) for the calculations, distance $r_h = d_h / 2$, where the specific heat flow density $q_r(r_h)$ is $0.05q_{\max}$, is assumed to be the radius of the heating spot. Hence, it follows that the heating spot diameter can be calculated from the following expression:

$$d_h = \frac{3.46}{\sqrt{k}}. \qquad (3)$$

The argon-arc welding (AAW) process is used mainly for modelling of the solidification cracking conditions [9–13]. Thermal characteristics of low-amperage arcs in AAW have approximate quantitative values [8–10] which greatly distort the picture of a real process in mathematical modelling. This made it necessary to experimentally study dependence of the effective thermal power $q_e$, efficiency $\eta_i$ of heating of an item, heat flow concentration factor $k$ and size of the weld upon the parameters of AAW using low-amperage arcs.

The purpose of this study was to investigate energy characteristics of the low-amperage arcs in AAW.

The effective thermal power of the arcs in AAW was determined by the method of immersion calorimetry, and the heat flow concentration factor was estimated by two-section flow calorimetry using experimental devices described in [14]. The effective efficiency of heating of a workpiece was calculated as a ratio of the effective power of the heat flow to consumed electric power of the arc. Experiments were conducted by welding stainless steel plates using the AAW torch and power supply VSVU 315. Arc current $I_a$ and arc length $l_a$ ranged from 20 to 120 A and from 2.5 to 10 mm, respectively. Tungsten electrode had 3 mm diameter, 30° pointing angle, and 0.4 mm tip face radius. Argon was used as a shielding gas. Gas flow rate $Q$ and welding speed $v_w$ were 10 l / min and 8 m / h, respectively, in all experiments. Welding was performed on the Kh20N16AG6 type stainless steel plates measur-

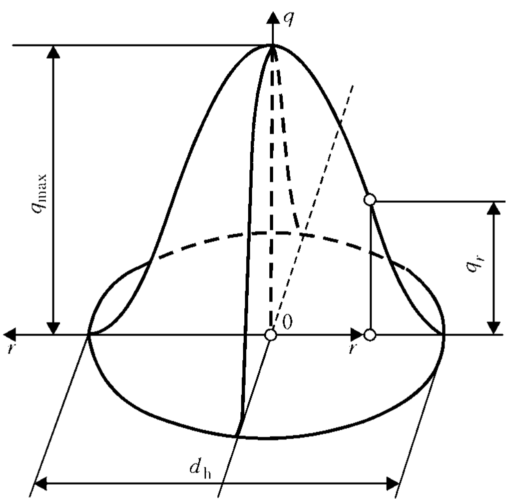


**Figure 1.** Schematic of distribution of the arc heat flow [7]

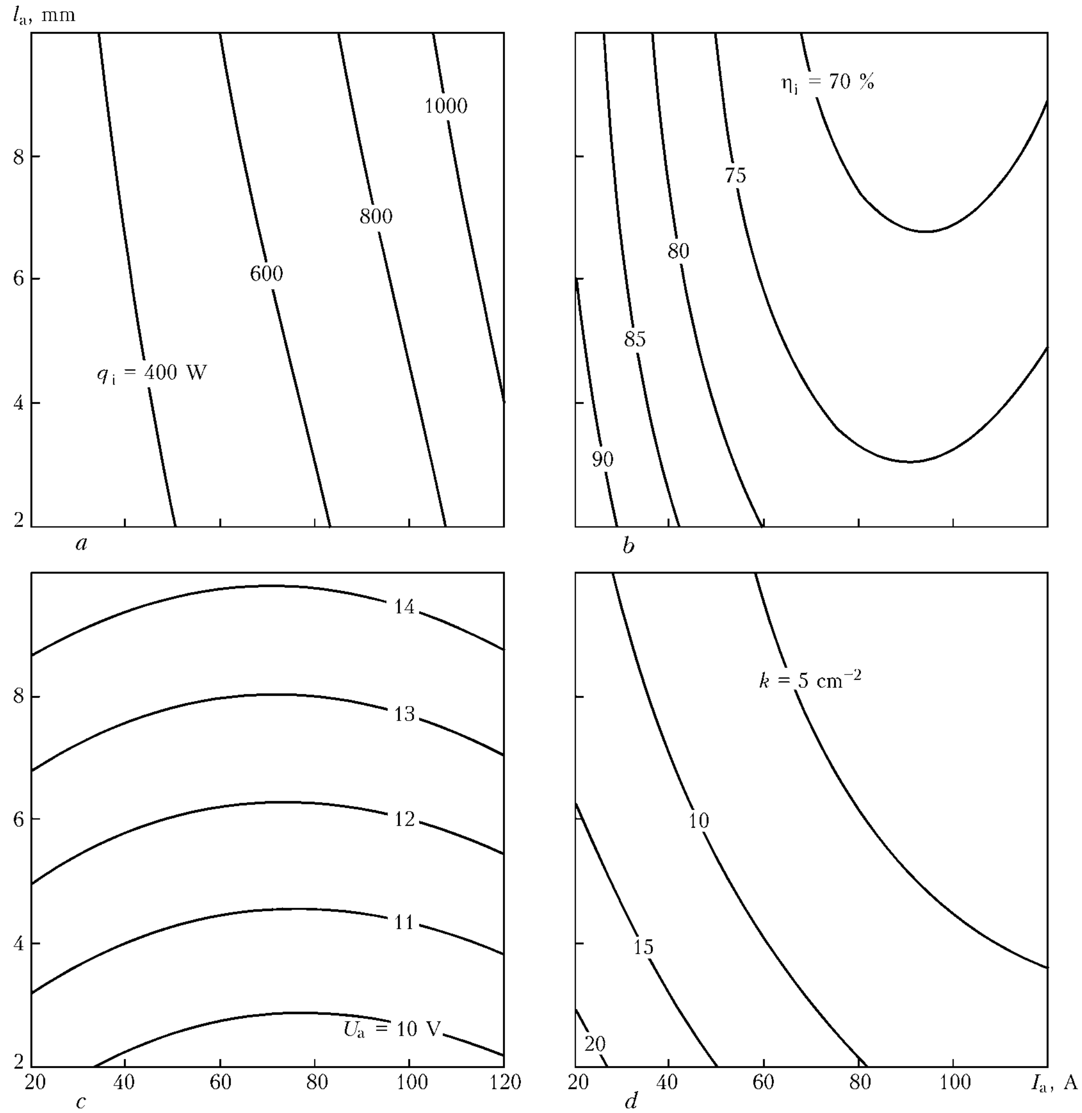


**Figure 2.** Effective thermal power of the arc $q_i$ (*a*) and effective efficiency $\eta_i$ of heating of an item (*b*), arc voltage $U_a$ (*c*) and heat flow concentration factor $k$ (*d*) versus arc current $I_a$ and arc length $l_a$

ing 90 × 40 mm and $\delta$ = 5 mm thick. After welding the plates were cut and etched in acid solution. Weld width $B_w$ and penetration depth $h_p$ were measured using a toolmaker's microscope with a ×10 magnification. Each experiment was repeated not less than four times. Statistical processing of the experimental data resulted in the following regression equations having the form of second-degree polynomials:

$$\eta_i(I_a, l_a) = 108.227 - 0.636 I_a - 1.396 l_a - 6.142\cdot 10^{-3} I_a l_a + 3.618\cdot 10^{-4} I_a^2 + 0.063 l_a^2 \ \ [\%]; \quad (4)$$

$$k(I_a, l_a) = 31.659 - 0.314 I_a - 2.342 l_a + 5.4\cdot 10^{-3} I_a l_a + 1.114\cdot 10^{-3} I_a^2 + 0.081 l_a^2 \ \ [cm^{-2}]; \quad (5)$$

$$h_p(I_a, l_a) = -1.069 + 0.054 I_a - 0.6 l_a + 1.278\cdot 10^{-3} I_a l_a - 1.1\cdot 10^{-4} I_a^2 + 0.017 l_a^2 \ \ [mm]. \quad (6)$$

As seen from the curves in Figure 2 plotted on the basis of the regression equations, the effective thermal power of the arc depends primarily upon the arc current. The effective efficiency of heating of a workpiece in AAW within the arc current range under consideration is 70–90 %, which is 2–10 % higher than in plasma-arc cladding using powder (PAC) (Figure 3, *a*) [6, 14]. It decreases with increase in the arc current, which is associated with growth of the heat losses caused by increase in cross section of the arc. Values of the heat flow concentration factor $k$ also markedly decrease with increase in the arc current, in contrast to the constricted arcs of cladding plasmatrons, where $k$ is almost independent of the current (Figure 3, *b*), and its values range from 4 to 18 cm$^{-2}$.

Change in the arc length from 2 to 10 mm has an insignificant effect on thermal characteristics of the arc. Increase in $l_a$ is accompanied by some increase in $q_i$ and a simultaneous decrease in $\eta_i$ and $k$, this dependence for the latter being much more pronounced with increase in $I_a$ (see Figure 2, *a*, *b*, *d*).

As shown by the investigation results presented in this study, as well as in [6, 14], at characteristic deposition process parameters in the case of AAW using filler wire ($l_a$ = 5 mm) and PAC ($l_a$ = 7.5 mm, diameters of the plasma and shaping nozzles are equal to 3.5 and 9.0 mm), the arcs in AAW are characterised by a higher heat flow concentration compared with PAC (Figures 3 and 4), which provides deep penetration and substantial concentration of stresses in the base metal.

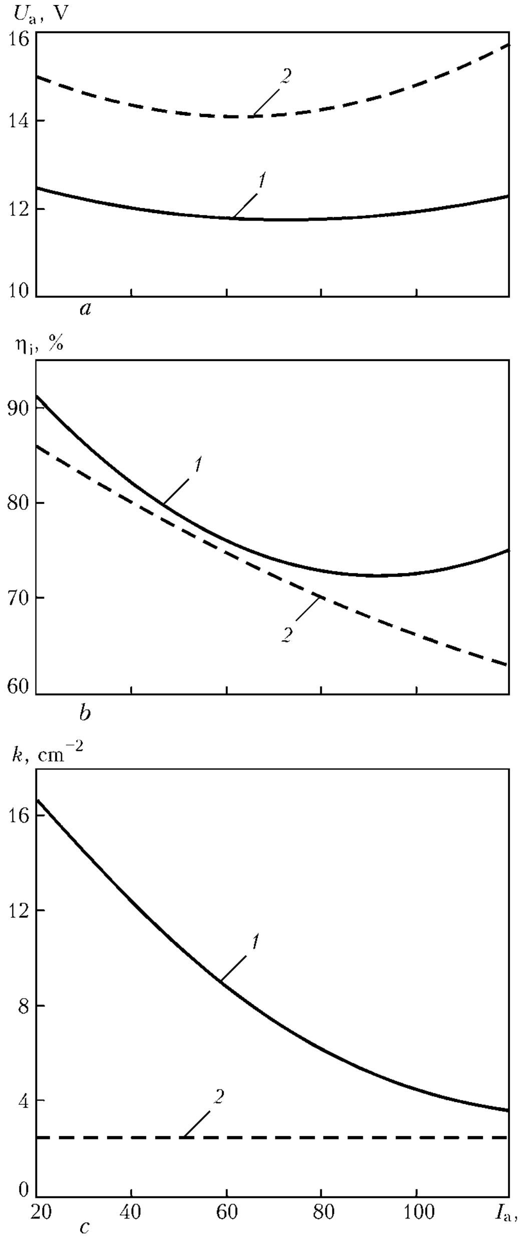


**Figure 3.** Arc voltage $U_a$ (*a*), effective efficiency $\eta_i$ of heating of an item (*b*) and heat flow concentration factor *k* (*c*) versus arc current $I_a$ in AAW (*1*) and PAC (*2*) [14]

Temperature fields in the base metal at the assigned process parameters can be analytically determined with a sufficient accuracy using the obtained values of the effective efficiency of heating of a workpiece and heat flow concentration factor. Figure 5 shows thermal cycles at a point on the weld axis at the plate surface heated to the same temperature in AAW ($q_i$ = = 580 W) and PAC ($q_i$ = 1000 W), determined by the method of a normally distributed heat source. The temperature in the initial period grows twice as fast to a maximal value in AAW. Dwelling of metal under the maximal temperature conditions in PAC is much longer. The final period of metal cooling in PAC is characteristic of a lower cooling rate, compared with AAW. As seen from Figure 6 (mathematical modelling of welding of stainless steel with $T_{melt} \approx 1420$ °C), the gradient of temperatures in HAZ grows with increase in the heat flow concentration, which is a negative factor for weldability of heat-resistant materials,

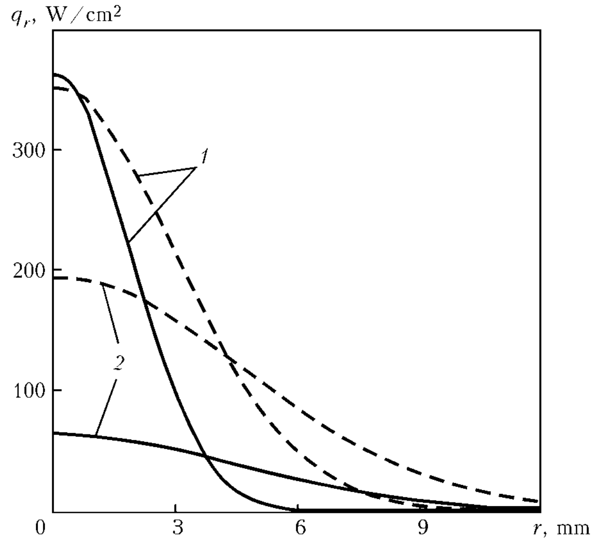


**Figure 4.** Distribution of specific heat flow of the arc, $q_r$, on radius *r* of the heating spot in AAW (*1*) and PAC (*2*) at $I_a$ = 30 (solid curves) and 100 A (dashed curves)

as a high stress gradient leads to hot cracking. Increase in the heat flow concentration favours growth of the weld penetration depth and width, although in this case the arc length and, hence, stress decrease, which was experimentally proved (Figure 7).

In addition to the heat flow concentration, the heat input also affects geometric parameters of the weld [12] (Figure 8, *a*, *b*). For example, the weld width and penetration depth on stainless steel decrease with increase in the arc length at a heat input of up to 700 J / mm. The drooping curve becomes less sloping with increase in the heat input. At a heat input of more than 700 J / mm, a change in the weld width with growth of the arc length (arc voltage) is characterised by a monotonously rising curve (Figure 8, *a*).

Therefore, the main parameters which allow a wide-range control of the weld pool formation and strained state of the HAZ metal are the concentration of the heat flow and heat input.

In welding of heat-resistant alloys, to increase resistance to solidification cracking the welding process should be performed with a low heat input and low concentration of the heat flow, thus providing a low temperature gradient in HAZ and minimal heat input to the base metal.

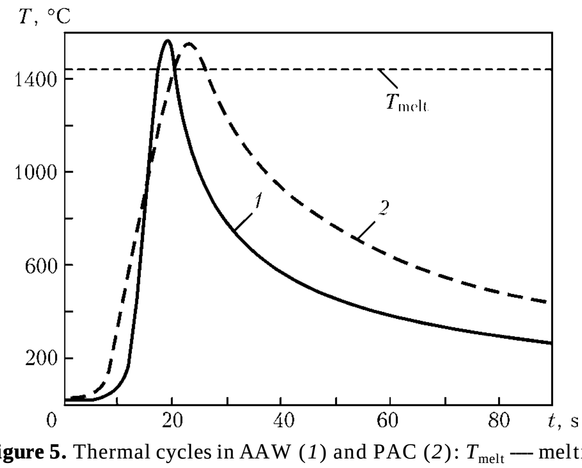


**Figure 5.** Thermal cycles in AAW (*1*) and PAC (*2*): $T_{melt}$ — melting temperature

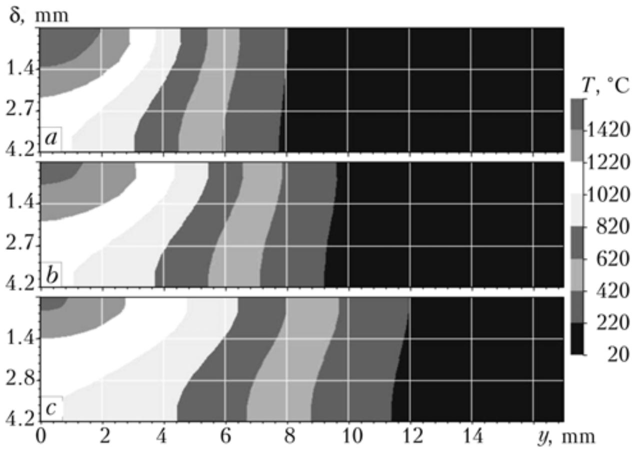


**Figure 6.** Temperature field and weld pool across the section of a plate (mathematical calculation was made by the method of a normally distributed heat source [8]): *a* — $l_a$ = 2.5; *b* — 6.0; *c* — 10.0 mm

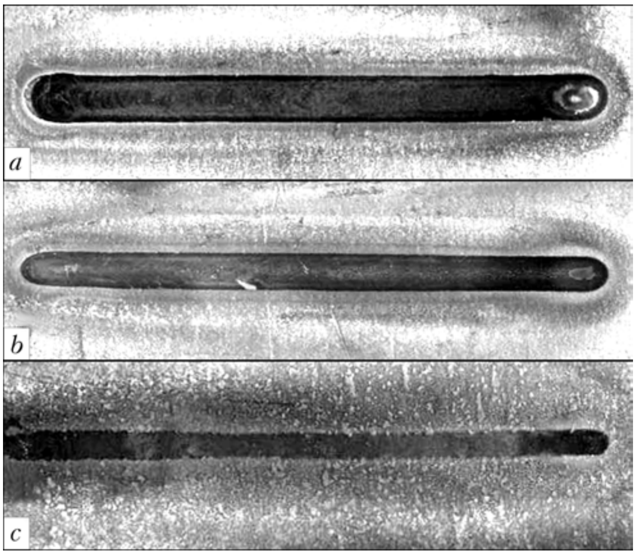


**Figure 7.** Appearance of deposited metal and HAZ on stainless steel plate (*a*--*c* — see Figure 6)

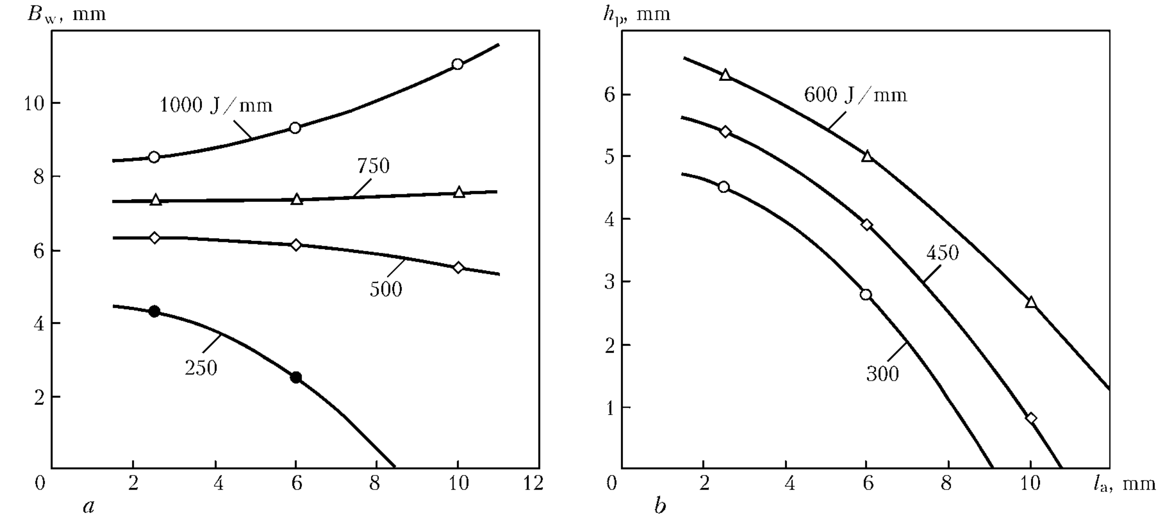


**Figure 8.** Weld width $B_w$ (*a*) and penetration depth $h_p$ (*b*) versus arc length and welding heat input

## CONCLUSIONS

1. In AAW with a low-amperage arc the effective efficiency of heating of a workpiece ranges from 70 to 85 % at $I_a$ = 20--120 A.

2. The heat flow concentration factor in AAW depends to a considerable degree upon the arc current and length, and ranges from 3 to 20 cm$^{-2}$.

3. The weld width and penetration depth at heat input of up to 700 J / mm decrease with increase in the arc length. Further increase in heat input leads to increase in the weld width.

4. At characteristic deposition parameters the low-amperage arcs in AAW are characterised by an increased (3–5 times) concentration of the heat flow, compared with arcs in PAC.


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# PECULIARITIES OF THE PROCESS OF MICROPLASMA WIRE SPRAYING

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Technology for microplasma spraying of coatings provides a small size of the spraying spot (several millimetres), is characterised by small dimensions of the equipment involved, and a low level of noise. Microplasmatron with a special wire feed device was designed to use wire as a source material for spraying. Structure and properties of coatings produced by the microplasma method using Mo, W, Ti, Ni--Cr, Al, brass and steel wire were investigated. The investigation results prove the possibility of applying this method for production of dense coatings with a spraying spot ranging from 3 to 8 mm.

*Keywords: microplasma spraying, plasmatron, wire feed mechanism, wire spraying, wire materials, copper, tungsten, coating, microstructure*

One of the thermal spraying methods is spraying of coatings using wire as a source material. Wire spraying can be performed by the plasma, flame or arc metallising methods. The plasma method can be realised using the wire–anode or neutral wire schemes. Both solid wires of metals and alloys (copper, aluminium, iron, titanium, nickel, chromium, molybdenum, tungsten, different alloys, etc.) and flux-cored wires of different compositions can be used as source materials [1].

The process of plasma wire spraying is characterised by the following peculiarities:

• all particles formed in wire atomisation are droplets of the melt;

• consumption of a spraying material is characterised by accuracy and stability;

• high productivity of the spraying process is provided by the wire–anode scheme.

Owing to the above peculiarities, coatings produced by plasma wire spraying have a dense lamellar structure, low porosity, high strength and hardness, low scatter of microhardness values, and low oxide content [2, 3].

Up to now, wire spraying has been performed using high-power plasmatrons. For example, the UPU-8M unit is fitted with a 40 kW plasmatron, which can perform spraying using 0.8--1.2 mm diameter wire. In the case of deposition of coatings on parts with a wall thickness less than 1 mm, this causes a risk of local overheating and buckling of the part, while the case of spraying on narrow stiffeners or strips is characterised by high losses of a spraying material.

The E.O. Paton Electric Welding Institute has recently developed a new version of plasma spraying –– microplasma spraying [4–6]. The MPS-004 system comprises a power supply with control panel, plasmatron and special powder feeder. Design of the plasmatron is covered by a patent of Ukraine [7]. Its working parameters provide formation of a laminar plasma jet, thus resulting in a number of peculiarities of the microplasma spraying process:

• decrease in the spraying spot size to 1--5 mm (expansion angle of the laminar plasma jet is only 2--6°, instead of 10–18° for turbulent plasma jets, and the nozzle diameter is 0.8–2.0 mm);

• possibility of depositing coatings on small-size and thin-walled parts without local overheating and buckling (low thermal power of the microplasma jet allows heating of the substrate to be decreased);

• low level of noise of the laminar plasma jet (30–50 dB).

**Specifications of the MPS-004 system for microplasma spraying of coatings**

| | |
|---|---|
| Working gas | argon |
| Shielding gas | argon |
| Power, kW | $\leq 3.0$ |
| Current, A | 10–50 |
| Voltage, V | $\leq 60$ |
| Plasma gas flow rate, l / h | 10–250 |
| Shielding gas flow rate, l / h | 60–400 |
| Dimensions, mm | 390 × 225 × 205 |
| Weight, kg | 14 |

The E.O. Paton Electric Welding Institute developed the microplasma spraying method for deposition of coatings using wire materials [8, 9]. It allows realisation of advantages of the wire spraying method under conditions of the microplasma jet. For this, the Institute made a specialised attachment to the plasmatron, which is part of the microplasma spraying system MPS-004 (Figure 1). The attachment is a compact mechanism for feeding a wire to an inter-electrode region of the plasma jet. The feeding mechanism is made with a possibility of regulating the wire feed speed by varying the quantity of revolutions at the electric motor shaft. The attachment has a small weight and is mounted on the location of a powder feeder.

Design peculiarity of the plasmatron is a remote anode. A shielding gas, i.e. argon, is fed to stabilise the plasma jet and shield the anode. The presence of a shielding atmosphere within the wire melting zone allows the degree of oxidation of a spraying material to be decreased, which is particularly important for

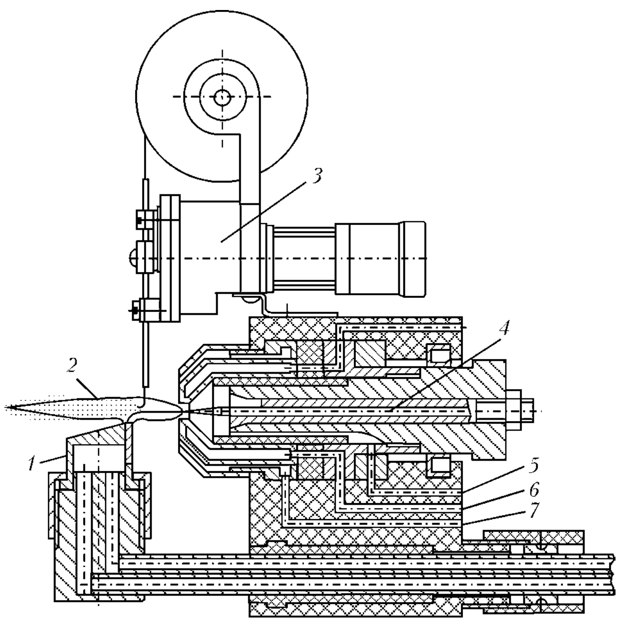


**Figure 1.** Diagram of the microplasma wire spraying plasmatron: *1* — anode; *2* — plasma jet; *3* — wire feed mechanism; *4* — cathode; *5*–*7* — channels for feeding the plasma gas, water and shielding gas, respectively

spraying of reactive metals, such as tungsten, molybdenum, titanium, zinc and brass.

Analysis of condition of detachment of a droplet in plasma wire spraying shows that the droplet detaches when the force induced by the jet, $F_j$, is equal to or higher than the surface tension force $F_{s.t}$ that holds it at the wire tip [3]. This condition can be described by the following equality:

$$F_j \geq F_{s.t}. \quad (1)$$

The droplet formed in a flow is affected by the drag force of the jet, which can be calculated from the following relationship:

$$F_j = 0.5 c_x \rho v^2 S, \quad (2)$$

where $c_x$ is the coefficient of the drag force of the jet, which depends upon the Reynolds number; $\rho$ is the density of the jet gas; $\nu$ is the plasma jet velocity; and $S$ is the area of a midlength section of the melt droplet.

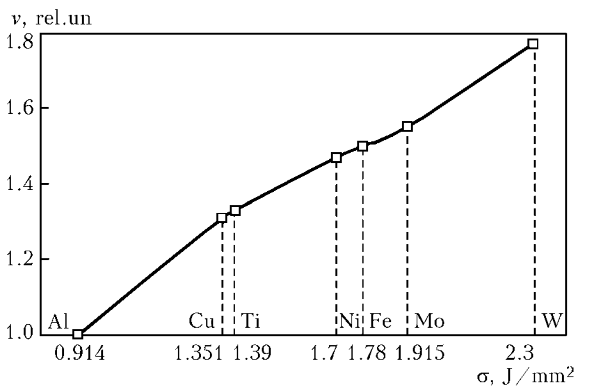


**Figure 2.** Required plasma jet velocity versus surface tension for different materials

As seen from formula (2), $F_j$ depends upon the flow velocity $\nu$. The force that holds the droplet at the wire tip, $F_{s.t}$, is proportional to wire diameter $d$ and surface tension force $\sigma$ of a given metal:

$$F_{s.t} = \pi d \sigma. \quad (3)$$

For microplasma wire spraying ($F_j > F_{s.t}$) it is necessary to provide a higher plasma jet velocity than in the case of powder spraying. In turn, the plasma jet velocity depends upon the nozzle diameter, plasma gas flow rate and plasmatron power. The plasma gas flow rate in wire spraying is 1.5–2 times higher than in spraying of powder materials [5, 6].

Temperature of the argon plasma jet in microplasma spraying is higher than that of a similar jet formed by conventional plasmatrons as a result of a higher power per unit flow rate of the plasma gas. The high temperature of the jet makes it possible to decrease the surface tension force of a detaching droplet due to overheating of the melt. The force holding the droplet at the wire tip, $F_{s.t}$, can also be decreased by decreasing the spraying wire diameter $d$. In this connection, in the case of refractory metal wire, it is recommended to select the wire of a smaller diameter. For example, for spraying a tungsten wire ($\sigma$ = = 2.3 J / m$^2$), its diameter should be 0.2 mm, although the steel ($\sigma$ = 1.78 J / m$^2$) and copper ($\sigma$ = = 1.35 J / m$^2$) wires can be easily sprayed at a diameter of 0.3 mm.

While considering equality (1) at an assumption that $c_x$ and $\rho$ are constant, it is possible to find relationship between the relative jet velocity (for aluminium it is assumed to be 1) required for wire spraying, and surface tension of different materials (Figure 2). It can be seen that the spraying jet velocity should be much higher for the materials with a high value of surface tension (tungsten, molybdenum) than for the materials with a low value of surface tension (aluminium, copper, titanium).

Therefore, in microplasma wire spraying the favourable conditions result from a combination of the following factors: high specific power (high enthalpy of the plasma jet), high temperature of the plasma jet, high velocity of the plasma jet, and utilisation of small diameter wires (0.3 mm).

As established as a result of the conducted experiments, the stable process of spraying of a neutral wire by the microplasma jet, using the shaping nozzle with a diameter of 0.8 mm and plasma gas flow rate of up to 300 l / h, takes place in the case of the 0.2–0.4 mm diameter wire. The 0.2 mm wire is used for refractory materials, such as tungsten (3395 °C) and molybdenum (2620 °C), the 0.3 mm wire is used for steel, copper, nichrome, titanium and others, and the 0.4 mm wire is used for low-melting point metals, such as lead (327 °C), tin (232 °C), etc. The speed of feeding the wire to the inter-electrode region of the plasma jet depends upon the amount of heat required for melting a unit volume of the wire, i.e. upon its diameter and thermal-physical properties of its mate-

**Table 1.** Amount of heat required for melting 1 cm of wire with $d$ = 0.3 mm for different materials

| Material | Material density ρ, kg / m$^3$ | Melting temperature $T_{melt}$, K | Melting heat Ñ·10$^{-5}$, J / kg | Specific heat ñ·10$^{-3}$, J / (kg·K) | Amount of heat required to melt 1 cm of wire with $d$ = 0.3 mm $Q_{melt}$·10$^{-2}$, J |
|---|---|---|---|---|---|
| Fe | 7860 | 1811 | 2.48 | 0.45 | 5.173 |
| Cu | 8920 | 1353 | 2.06 | 0.38 | 3.838 |
| Al | 2710 | 933 | 4.01 | 0.91 | 1.884 |
| Ni | 8960 | 1728 | 2.98 | 0.46 | 6.068 |
| Ti | 4500 | 1941 | 3.59 | 0.52 | 3.868 |
| Mo | 10200 | 2890 | 3.82 | 0.25 | 7.435 |
| W | 19230 | 3660 | 1.92 | 0.13 | 8.559 |

**Table 2.** Parameters of microplasma wire spraying depending upon the wire material (experimental data)

| Parameter | Wire material | | | | | | |
|---|---|---|---|---|---|---|---|
| | Steel | Cu | Ni–Cr | Brass | Ti | Mo | W |
| Wire diameter, mm | 0.3 | 0.3 | 0.3 | 0.3 | 0.3 | 0.2 | 0.2 |
| Wire feed speed, m / min | 5 | 6 | 4.5 | 5.5 | 5.5 | 8 | 9 |
| Voltage, V | 30 | 32 | 32 | 25 | 25 | 32 | 32 |
| Current, A | 45 | 48 | 48 | 20 | 30 | 48 | 50 |
| Specific power, W / l | 13.5 | 17.1 | 15.4 | 3.3 | 5 | 14 | 13.3 |
| Plasma gas flow rate, l / h | 100 | 90 | 100 | 150 | 150 | 110 | 120 |
| Shielding gas flow rate, l / h | 180 | 200 | 200 | 400 | 350 | 300 | 300 |
| Spraying material consumption, g / min | 2.75 | 3.7 | 3.3 | 2.7 | 1.6 | 2.6 | 2.2 |

rial. The amount of heat required to melt 1 cm of the wire can be determined from the following formula:

$$Q_{melt} = [c(T_{melt} - Ò_0) + C]\rho 0.01\pi d_w^2/4. \quad (4)$$

Considering properties of the materials used, as well as the fact that $T_0$ = 293 K, the values of $Q_{melt}$ can be found for each of the materials.

The calculation results are given in Table 1. It can be seen from the Table that for the 0.3 mm diameter copper and titanium wires ($Q_{melt} \approx 3.8 \cdot 10^2$ J) the optimal spraying conditions can be achieved at a higher wire feed speed, compared with steel and nichrome wires of the same diameter ($Q_{melt} > 5 \cdot 10^2$ J). It was established that to ensure the stable process of microplasma spraying of molybdenum and tungsten coatings, it is necessary to use the 0.2 mm wire and increased values of current and gas flow rates. In this case, the wire feed speed amounted to 10 m / min, which was 3–5 times higher than with the traditional methods of plasma spraying using wire materials [10, 11].

To determine the particle size composition of powders produced by microplasma jet atomisation of wires

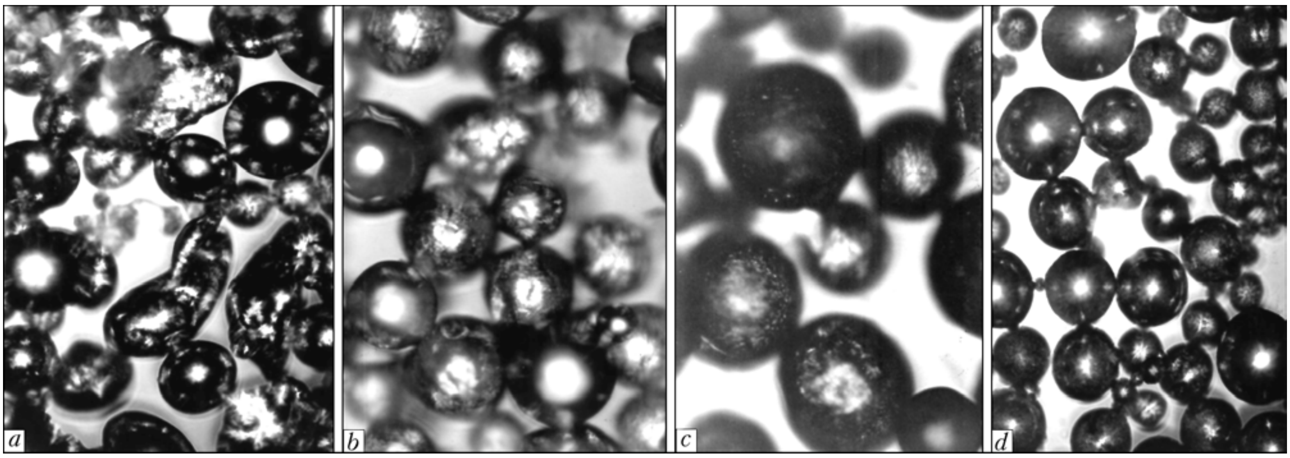


**Figure 3.** Appearance of powder particles produced by microplasma jet atomisation of wire: $a$ — tungsten, $b$ — copper; $c$ — nichrome; $d$ — steel (0.65 % C, 1 % Mn) (×200)

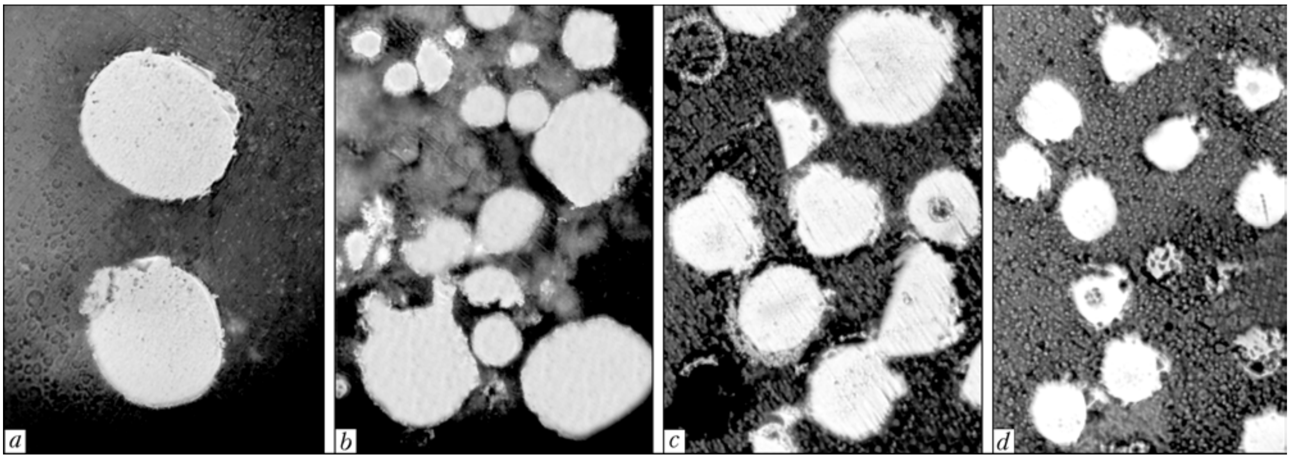

Figure 4. Microstructure of cross section of powder particles produced by microplasma jet atomisation of wire: *a* — tungsten, ×400; *b* — copper, ×400; *c* — nichrome, ×250; *d* — steel (0.65 % C, 1 % Mn), ×200

**Table 3.** Particle size composition of powders produced by microplasma jet atomisation of wire

| Wire material | Size of powder particles, μm | | | | | |
|---|---|---|---|---|---|---|
| | 0–20 | 20–45 | 45–63 | 63–80 | 80–100 | 100–160 |
| Copper | 15 | 30 | 40 | 10 | 5 | -- |
| Nichrome | 10 | 15 | 50 | 15 | 10 | -- |
| Steel (0.65 % C, 1 % Mn) | 29 | 15 | 20 | 15 | 10 | 1 |
| Tungsten | 10 | 10 | 15 | 20 | 30 | 15 |

and study the shape of particles of the atomisation products of different materials, the powder produced was collected in water. The wire atomisation parameters correspond to those given in Table 2. All powders were collected in a water filled pool at a distance of 300 mm from the atomisation point and at an angle of 30° to the water surface. After removal of water, the powders were dried in a drying cabinet (Figure 3). The particle size composition of powders was determined using a metallography microscope (Figure 4). Results of examination of the atomisation products of different materials are given in Table 3. For example, a particle size of 45--63 μm is dominant for the nichrome particles, the quantity of the rest of the particles symmetrically decreasing. In the case of the steel powder (0.65 % C, 1 % Mn) the particles 0--20 μm in size are dominant, the quantity of the rest of the particles decreasing with increase in size of the particles. In atomisation of the tungsten wire, the quantity of particles of different sizes is approximately the same, some increase being observed in a range of 63--100 μm.

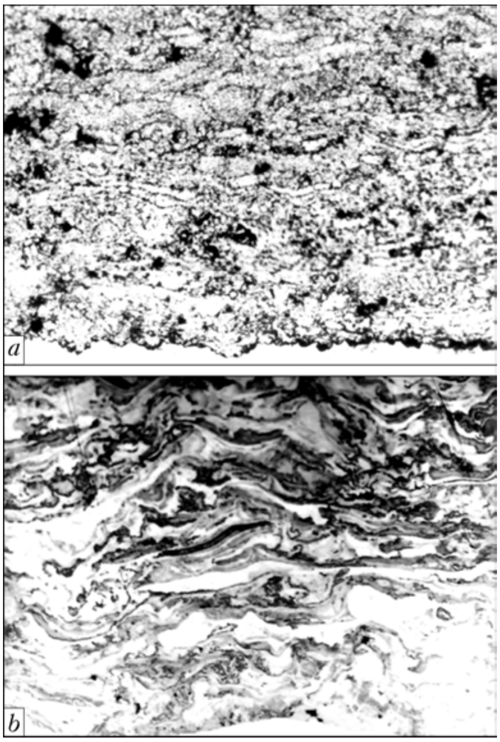

**Figure 5.** Microstructure of coatings produced by microplasma spraying using tungsten (etched) (*a*) and copper (unetched) (*b*) wires (×200)

Dependencies of the plasma and shielding gas flow rates, voltage, current, wire diameter, and feed speed upon the wire material were established as a result of experiments on determination of optimal conditions of microplasma spraying of coatings using wires of different materials and different diameters. The experimental results are given in Table 2.

Microstructure of plasma coatings produced by spraying of tungsten and copper wires is shown in Figure 5.

Tungsten coating 1.0 mm thick has a sufficiently dense structure, without spallings and cracks, and is characterised by a good adhesion to the substrate. It has a lamellar structure. The internal structure of lamellae is fine and columnar in some locations. The lamellae are 5--10 μm thick. According to the data of X-ray phase analysis (Figure 6, *a*), the coating contains one phase —- tungsten. Microhardness of the tungsten coating is 1880–2060 MPa.

Copper coating (Figure 5, *b*) 0.5 mm thick has a dense structure, like the tungsten one, contains no spallings and cracks, and is characterised by a good adhesion to the substrate. Its structure is lamellar and heterogeneous. The lamellae are 5--30 μm thick. The main phase is copper (Figure 6, *b*). In addition, the coating contains inclusions $Cu_2O$ and regions of eutectic Cu–$Cu_2O$. Microhardness of the copper coating is 450–610 (copper), 870–910 ($Cu_2O$), and 1000–1100 MPa (eutectic Cu–$Cu_2O$).

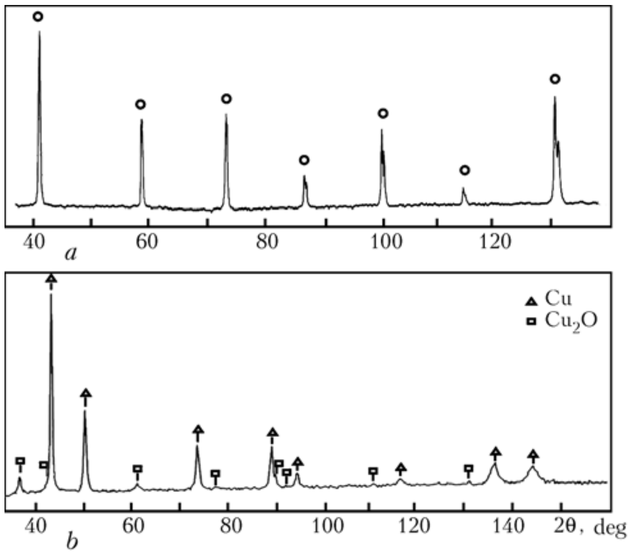


**Figure 6.** Phase composition of tungsten (*a*) and copper (*b*) coatings produced by microplasma wire spraying

## CONCLUSIONS

1. Factors affecting the process of microplasma spraying using wire materials include current, plasma gas flow rate, wire diameter and feed speed. Analysis of conditions of droplet detachment in microplasma wire spraying shows that in this case it is necessary to use small-diameter wires (0.2–0.4 mm) and increased wire feed speeds (5–10 m / min).

2. Flow rate of the plasma gas (argon) is related to wire diameter and surface tension of the metal melt. The required velocity of the plasma jet, at which the optimal conditions of formation of the melt droplet at the wire tip and subsequent splitting (dispersion) are ensured, was found to depend upon the surface tension of the melt of different metals.

3. In wire atomisation, the content of particles less than 63 μm in size in the case of copper is 85, steel and nichrome —- 75, and tungsten —- 35 wt.%. The particles have a shape close to spherical.

4. Analysis of microstructure of coatings shows that the microplasma coatings produced by wire spraying have a dense structure (with lamellae 5--30 μm thick) and a low degree of material oxidation, this being related to the application of the shielding gas.

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# INDIAN WELDING INDUSTRY: CURRENT SCENARIO AND PROSPECTS

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Considered are the state-of-the-art and future prospects of welding production in India, as well as the current status of international cooperation in the field of welding. The data on volumes of production and utilisation of steel and welding consumables are given. Main welding processes applied in India are reviewed. Prospects of development and application of power supplied are considered.

*Keywords: welding industry, steel consumption, consumption of welding consumables, welding processes, power sources, advanced technologies, collaboration areas*

**Steel consumption.** Growth of welding technology and the industry in India has been slower than desired, but the future looks bright. Size of the welding industry in any country has a direct relation with the consumption of the steel in the country. Global consumption of steel is likely to exceed 1000 million, i.e. 1 billion tons in 2005. Against this, the consumption of steel in India was of the order of 30.4 million tons in 2003--2004 and around 32.5 million tons in 2004--2005. With approximately one-sixth, i.e. 16 % of the global population living in India, the consumption of steel is around 3 % of the global consumption.

Per-capita consumption of steel in India is as low as around 30 kg, compared to around 450 kg in the developed countries, the global average being over 150 kg. It is noteworthy that per-capita consumption of steel in China in 2003 was 250 kg, which further went-up in 2004. Whereas all-India per-capita consumption is 30 kg, consumption in rural areas in India is 2 kg.

All this, when India has rich resources of iron ore and is one of the lowest cost producer of mild steel in the world. It is not for nothing that India exported 5.3 million tons, i.e. 15 % of its production in 2003–2004.

It is apparent from the above that there is only one way to go for steel consumption in India and that is up. As regards the expected growth rate, estimates vary between 6 to 7 % for the next few years. All steel majors in the country are at various stages of substantial enhancement of their capacity. The draft steel policy projects Indian steel making capacity to be in the region of 60 million tons by 2010 and 110 million tons by 2020. Ministry of Steel, Government of India quotes consumption of steel in the country in 2020 to be 100 million tons. Two of the world's largest steel producers, namely Mittal Steel Company and POSCO, are in the processes of starting steel production in India. For this article, we assume a growth rate of 6.5 %.

**Relevant factors.** *Construction*. A recently released study from Global Insight Incorporated of the USA, entitled «Global Construction Study 2003» measures and forecasts upto the year 2025 construction industry spending in 55 of the world's largest construction markets. The study forecasts moderate global growth at 5 % in construction investment up to 2012, with India growing at the highest rate amongst large economies at 9.2 %. This is even higher than China. According to the study, India and China, in that order, are expected to make exciting news by offering major opportunities and increase in revenue for construction companies. In India, government infrastructure initiatives for building roads, rail roads, bridges and power lines would be helping create an expected growth rate of 10.6 % over the next five years. The government has also planned projects for development and modernization of the ports on the East and the West coasts of India. If and when inter-linking of the rivers in India takes shape, one can imagine the magnitude of construction activity.

*Fabrication*. What about fabrication industry? Thanks to technological skills, lower labour and manufacturing costs, India is taking great leaps in fabrication industry. Many of the large fabricators, including Larsen&Toubro (L&T), Bharat Heavy Electricals Ltd. (BHEL), ISGEC, etc. are winning contracts for fabrication of pressure vessels, heat exchangers, columns, power plant equipment etc. in the global market, including China.

*Competitiveness*. One of the Europe's leading business schools IMD, Switzerland brings out the world competitiveness report every year. In the World Competitiveness Year Book (WCYB) for 2004, India has jumped a massive 16 ranks. It now stands at 34th up from the 50th position it held in 2003. This is the largest jump by any country among the 60 countries ranked annually in the WCYB. More over, this is the best rank for India till date. The marked improvement in India's position this time is because of a huge jump in three of the four parameters on which the competitiveness is measured. These are economic performance, business efficiency and government efficiency. As far as the fourth parameter, i.e. the infrastructure goes, India still remains close to the bottom. It was 58 last time and has moved up just one rank to 57.

*Others*. Within India, the Electricity Act 2003 has been enacted. This act is sure to revive investors' interest in power generation and distribution, which was in limbo for over a decade. The auto industry is witnessing a growth boom. The consumer durable industry has picked-up.

**Consumption pattern of welding consumables.** *Present consumption*. We would like to first make an estimation of the total weld metal deposited in India during the year 2003--2004 and the share of the major welding processes. Based on the actual consumption of steel as 30.4 million tons, the total weld metal deposited in 2003–2004, 0.5 % of steel has been 152,000 tons. Welding consumables industry in India is highly fragmented, with an approximate one-third of the total value shared by more than a hundred of manufacturers in the unorganized / small-scale sector. It is therefore extremely difficult to arrive at an accurate estimate of data regarding the share of various processes in the weld metal consumption. Still, based on various inputs, we would venture an estimated share of welding processes in 2003–2004 as SMAW — 75 %, GMAW — 17 %, SAW — 7 %, and GTAW and others — 1 %.

*Projections*. Consumption of steel in India is expected to grow at 6.5 %. The share of SMAW process will decrease and the share of GMAW process will increase as the industry progresses. Such transformations have been noticed in all developed and developing countries. For the purpose of this study, we are assuming that the share of SMAW will gradually go down from the present 75 to 65 %, and the share of GMAW will gradually go up from the present 17 to 27 % during the 10 years since 2003–2004. As observed even in developed countries, share of SAW will not change much from the present 7 %. Similarly, share of GTAW and other processes will remain around 1 %.

Based on the inputs / assumptions mentioned above, the current and projected share of the major welding processes and the consumables for the next 8 years is given in the Table (in tons).

*Usage of flux cored wires*. Rough estimates of flux cored wire consumption in 2003–2004 are about 3000 tons, catered to by both local and imported products. The applications cover joining as well as hard-facing / surfacing. Good quality welds produced by flux-cored wires is a big advantage. There are a few companies in India who manufacture flux-cored wires for joining and hard-facing applications. However, the high cost of flux-cored wires restricts its use to only those customers who have relatively high labour wage rate, for example, L&T, Thermax, BHEL etc. Resultantly, flux cored wires presently constitute only around 10 % of the total GMAW process and less than 2 % of the total weld metal deposited in the country. This compares poorly with some of the advanced countries, where as much as 25 % of the total weld metal gets deposited by flux cored and / or metal cored wires. If and when the price of flux-cored wires is brought down in the country, the share of FCAW process will find many more applications.

**Power sources.** In arc welding power sources, India is a land of contrasts. Use of antiquated bare welding transformers by roadside welders and cottage type welding works is a common sight. These energy guzzlers continue to be used at low-end applications. The organized fabrication and construction works have been gradually moving from motor generator sets to silicon diode rectifiers, which help in energy conservation. With the advances in power electronics, the

| Year | Parameter | SMAW | GMAW | SAW | GTAW and others | Total |
|---|---|---|---|---|---|---|
| 2003–2004 | Weld metal<br>Electrodes / wires | 114000<br>182400 | 25840<br>28420 | 10650<br>11700 | 1520<br>1670 | 152000<br>224190 |
| 2004–2005 | Weld metal<br>Electrodes / wires | 119880<br>191810 | 29160<br>32080 | 11340<br>12470 | 1620<br>1780 | 162000<br>238140 |
| 2005–2006 | Weld metal<br>Electrodes / wires | 125850<br>201360 | 32760<br>36040 | 12070<br>13280 | 1720<br>1890 | 172400<br>252570 |
| 2006–2007 | Weld metal<br>Electrodes / wires | 132190<br>211500 | 36720<br>40390 | 12850<br>14130 | 1840<br>2020 | 183600<br>268040 |
| 2007–2008 | Weld metal<br>Electrodes / wires | 138820<br>222110 | 41050<br>45150 | 13680<br>15050 | 1950<br>2140 | 195500<br>284450 |
| 2008–2009 | Weld metal<br>Electrodes / wires | 145600<br>232960 | 45760<br>50340 | 14560<br>16020 | 2080<br>2290 | 208000<br>301610 |
| 2009–2010 | Weld metal<br>Electrodes / wires | 153060<br>244900 | 51010<br>56110 | 15530<br>17080 | 2200<br>2420 | 221800<br>320510 |
| 2010–2011 | Weld metal<br>Electrodes / wires | 160620<br>256990 | 56690<br>62360 | 16530<br>18180 | 2360<br>2600 | 236200<br>340130 |
| 2011–2012 | Weld metal<br>Electrodes / wires | 168570<br>269710 | 62900<br>69190 | 17610<br>19370 | 2520<br>2770 | 251600<br>361040 |
| 2012–2013 | Weld metal<br>Electrodes / wires | 176820<br>282910 | 69650<br>76610 | 18750<br>20630 | 2680<br>2950 | 267900<br>383100 |

cost of silicon controlled rectifiers (thyristors) is much lower than the cost of generators. Although inverters have to be largely imported by India at present, the population of inverter power sources has been increasing. Since inverters reduce energy consumption by as much as 30–40 % and since the cost of energy in India is amongst the highest in the world, logically the use of inverters should multiply in India. Easy availability of inverter power sources and adequate service facilities would go a long way in this direction.

**Advanced processes**. India has started adopting and using processes like EBW, laser welding, cutting and brazing, hybrid laser + arc welding, water jet cutting, twin wire narrow gap submerged-arc welding, automated plasma transferred-arc welding, tandem pulsed GMAW, plasma arc welding, robot welding, variable polarity plasma arc welding and friction stir welding. These advanced processes are finding applications in aerospace components, high pressure boiler fabrication, atomic energy establishments, automobiles, valves and so on. These moves are expected to gather momentum as India shapes to improve its competitiveness in the global fabrication and construction markets. Economics of scale and the need to meet compressed delivery deadlines would act as the drivers for adopting advanced, high productivity and reliable welding processes.

**Collaboration possibilities in India.** India offers a host of strengths, some of which can be highlighted as:

• high educational standard of engineers, technicians and managers;

• good knowledge of English language and information technology;

• good skills of craftsmen and their willingness to upgrade;

• low cost of manpower across the board;

• sustained growth in low-end and medium level technology applications and virtually un-tapped in high-end welding processes;

• good knowledge of domestic market with Indian companies;

• fair and reliable judicial system;

• no governmental policy barriers, and welding related market is open to direct foreign investments.

Some leading welding companies like ESAB, Lincoln Electric, Oerlikon, Migatronic etc. already have successful presence in India.

India offers great opportunity as a manufacturing base not only for domestic consumption but also for exports.

Proven technology, possibly with capital investment, coupled with Indian strengths should prove perfect fits.

**Possible areas of collaboration:**

• India has a few institutions, which are engaged in applied research in welding technology. Examples are Welding Research Institute at Trichurapalli, many of the Indian institutes of technology, and a few engineering colleges. In addition, leading welding consumable and equipment manufacturers have their independent research and development centers. International research institutes and welding companies can utilize the talent available in such R&D establishments.

• Most of the Indian welding electrode manufacturers have established technology for meeting AWS and other international standard requirements. However, more advanced welding electrodes with very low impurities and for meeting the exacting requirements for high pressure / high-temperature power plant, refineries, petro-chemical applications are being imported. International welding companies, preferably with Indian partners, can establish their brands in India.

• As mentioned earlier, there is a definite scope for increasing the usage of flux-cored / metal-cored wires in India. The scope exists both for joining and surfacing / reclamation applications.

• In the area of welding power sources and equipment, there is tremendous scope for low-cost automation as well as inverters.

In general, the time appears right for leading international companies to establish their foothold in India in order to participate in the consistently growing Indian welding market.

# COMPUTER SYSTEM OF ELECTRON BEAM AND LASER WELDING MODELING

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System of engineering computer analysis of laser, electron beam and hybrid welding LaserCAD is described, which allows predicting the weld pool shape and dimensions, structure and properties of weld and HAZ metal. The system is based on physically adequate numerical-analytical models of laser and electron beam treatment processes, contains databases on materials and equipment, and has user-friendly interface.

*K e y w o r d s : laser welding, electron beam welding, laser-arc welding, mathematical models, engineering computer analysis*

State-of-the-art level of the industry development assumes active introduction and use of new methods of designing technological processes. Computer-aided design systems significantly accelerate process of developing new technologies and allow creating more and more complex technological objects.

At present many systems of different types exist — CAD–CAE–CAM, designed for using in various branches of industry and for research activity. However, there is a clear lack of these systems in the field of beam technologies.

State-of-the-art program of engineering designing should have necessary properties (use of adequate physical and mathematical models of technological processes, wide information basis, speed of operation, user-friendly dialogue interface) for ensuring computer-aided designing at all or separate stages of a technological process development.

The LaserCAD system is designed for solving various kinds of engineering tasks: calculating geometric parameters of an assumed joint, choice of optimum parameters of treatment conditions, selection of necessary equipment in agreement with necessary parameters, and choice of materials in accordance with assumed properties of a joint. It includes models of laser and electron beam welding with deep penetration, hybrid welding with deep penetration and surface melting, multilevel database on materials containing information on physical properties and chemical composition of the alloys, database on technological equipment, means of dialogue optimization of the laser, electron beam and hybrid welding processes, and user-friendly dialogue interface (Figure 1).

All mentioned models of beam welding represent a set of physically adequate models of processes, which proceed in various kinds of treatment. The model of laser welding with deep penetration includes solution of the task of heat transfer into metal in liquid and solid phases, hydrodynamics of the melt and metal vapor flow and formation of the laser-induced plasma, and interaction of the plasma with the beam.

Hybrid welding model, except processes characteristic of laser welding, takes also into account heating from an additional heat source (arc, plasma or light ones) in deep penetration and allow calculating temperature fields in surface melting on the basis of numeric solution of a 3D problem of non-stationary heat conductivity. The electron beam model includes solution of the problem of heat-and-mass transfer, gas dynamics and kinetics of metal vapor dissipation in

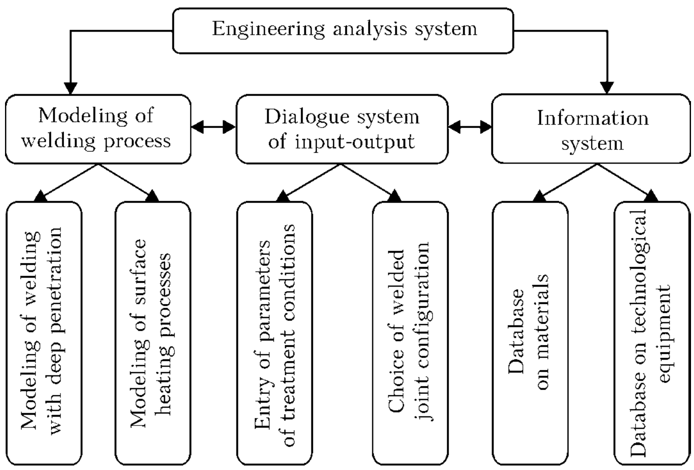


**Figure 1.** Flow chart of LaserCAD system for beam welding of metals

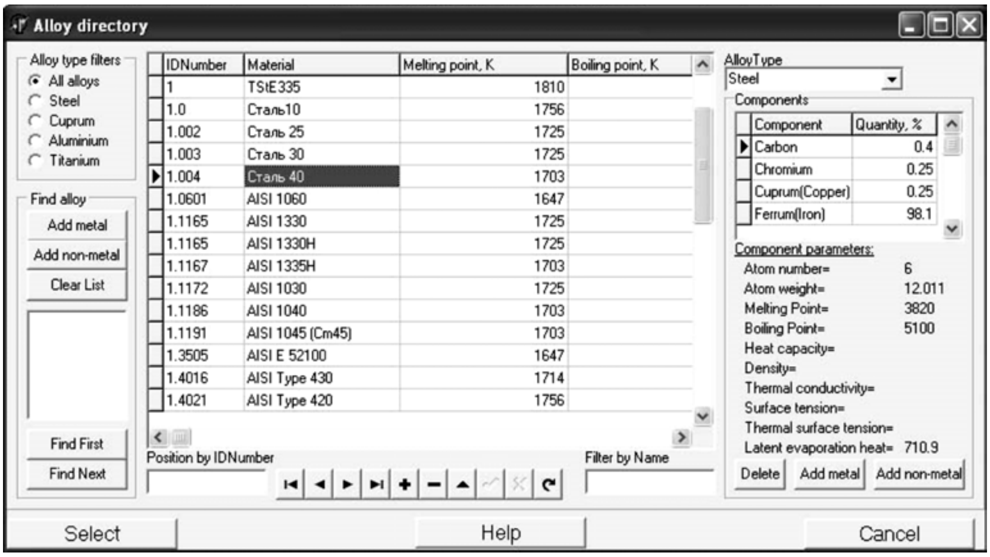


**Figure 2.** Database on properties of materials to be welded

vacuum and beam dissipation on metal vapors in welding with deep penetration.

The LaserCAD program contains multilevel complex reference system (Figure 2), which includes database on the material to be welded (information on physical properties of metals, non-metals and alloys, chemical composition thereof, and structure of steels obtained on the basis of thermal-kinetic diagrams of austenite decay). On the basis of these data and calculations obtained using mentioned models one can predict properties of the joint and the HAZ metal. The reference system also contains information on treatment equipment, which may be chosen according to the requirements to the treatment conditions or conditions of operation (Figure 3).

Databases are open for editing and updating, there is also possibility of direct entry into Internet.

The LaserCAD program has user-friendly interface, which makes it possible to work in the dialogue mode. A special function «Manual optimizer» facilitates and accelerates choice of the treatment mode for achievement of the necessary result (Figure 4).

The LaserCAD system is a means of engineering computer analysis CAE of the processes of metal welding by concentrated energy flows. It allows predicting size and shape of welded joints, structure and properties of the weld and HAZ metal, and analyzing changes, which take place in chemical composition of the weld metal during welding (Figure 5).

The system is organized according to the MDI-program and may operate in multi-window mode (Figure 6), which makes it possible to solve simultaneously several problems associated with comparison and analysis of the results.

The program is written in language Object Pascal in environment Delphi6 and developed in accordance with the main principles of construction of state-of-the-art systems CAD--CAM--CAE. The program operates in operational environment of the family Windows 95 / 98 / 2000 / XP; minimum system support is Pentium III / 700, 128 Mb RAM. Use of optimized algorithms allows achieving speed of the system operation necessary for application in the industrial prac-

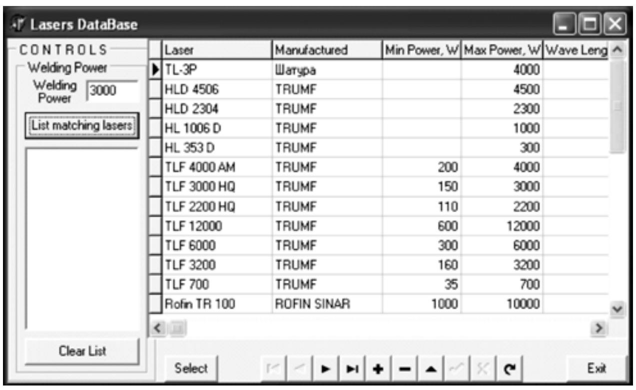


**Figure 3.** Database on laser equipment

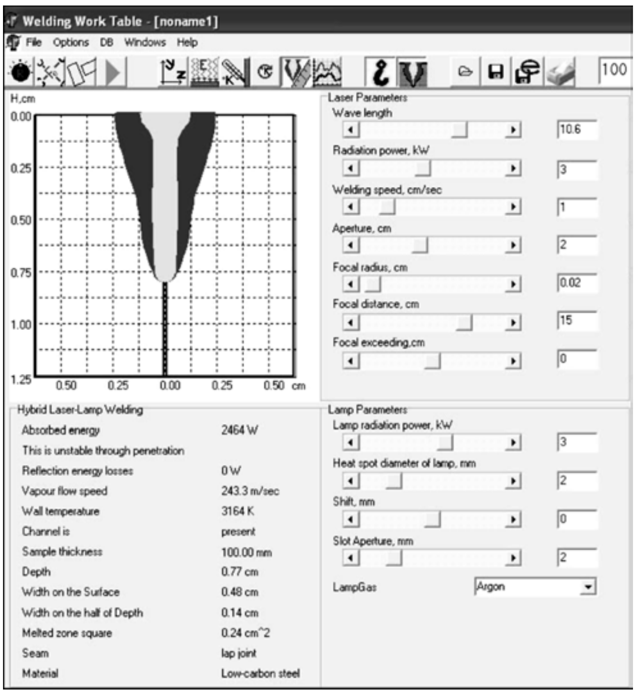


**Figure 4.** Dialogue optimizer for investigations

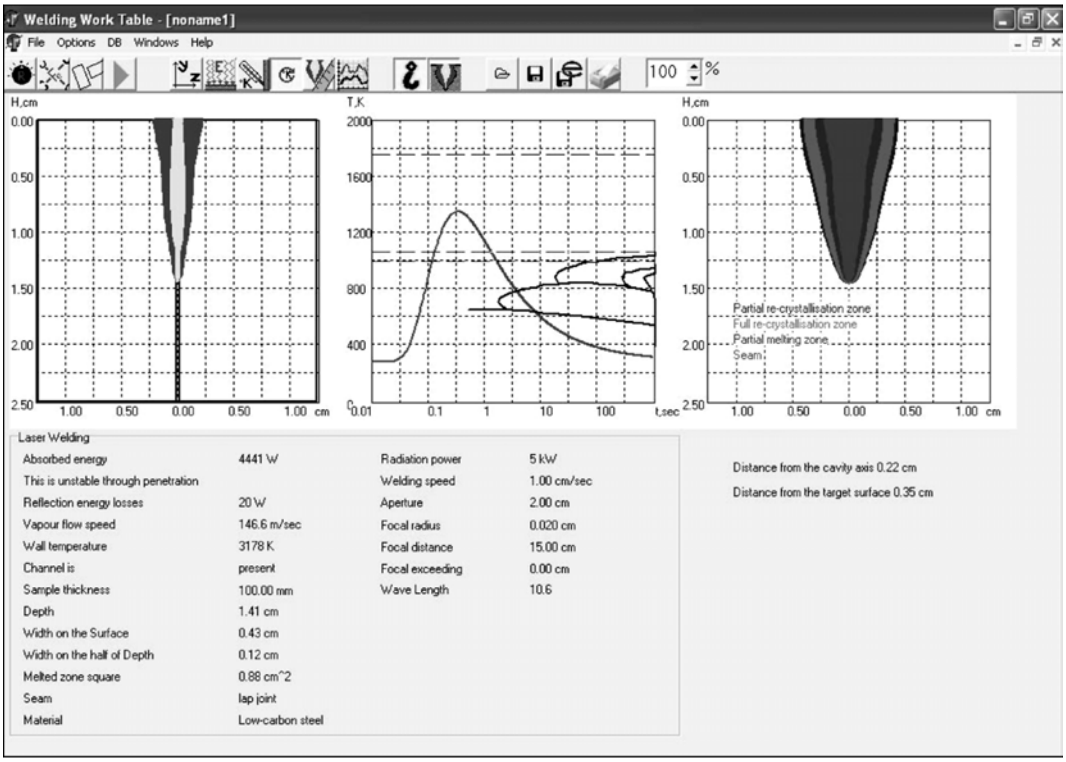


**Figure 5.** Analysis of welded joint structure of steel 30KhM in laser welding

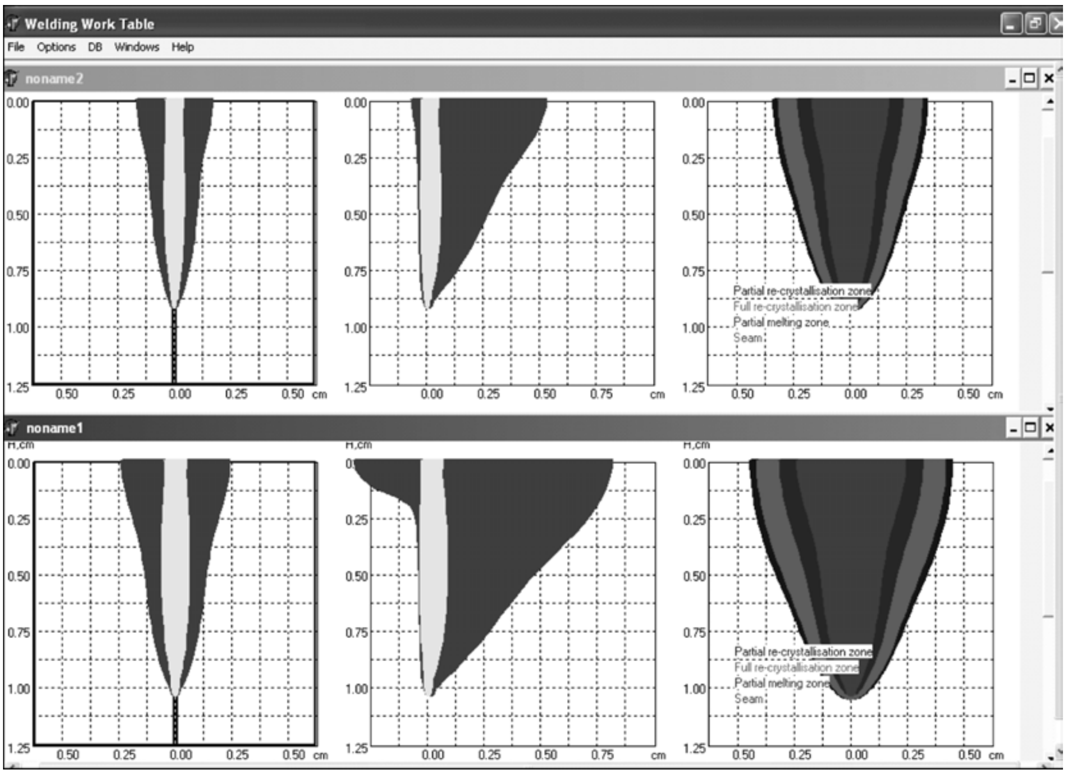


**Figure 6.** Example of simultaneous system operation in multi-window mode in laser and laser-light welding

tice; calculation of one option of the welding conditions takes maximum 1 min.

The LaserCAD system and its analogues are successfully used in Russia (in TsNIIRTK and TsNIITS, St.-Petersburg; N.E. Bauman MGTU, Moscow) and abroad in a number of institutes and companies (IWS, Weldaix, NMB, SLV Rostok). It may be used not just for designing technological processes, but also for training the personnel as a program for modeling beam technologies.

# APPLICATION OF OBJECT-ORIENTED DESIGN IN DEVELOPMENT OF WELDING TPACS

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Status of guidelines on development of welding automated control systems is analyzed. The possibility is shown of using object-oriented design in development of complex welding TPACS.

*Keywords: welding TPACS, object-oriented design, stages of AS development, inter-state, state and branch standards*

Efficient functioning of welding production requires for introducing at the enterprises advanced technologies and automated control systems. Due to the fact that development of a new automated technology for an enterprise is connected, as a rule, with temporary withdrawal of significant resources from the means of circulation, the need occurs to fulfill a project within minimum term with maximum correspondence of the final version of the welding technological process automated control system (TPACS) to the technical requirements.

As it is known, development of welding TPACS is usually performed by specialized organizations (scientific-engineering companies, scientific-research institutes, etc.) and only in rare cases by the enterprise itself, provided it has in its structure a respective sub-unit. In this connection of special importance is efficient interaction between a customer and the manufacturer. Existing guidelines on standardization of development and introduction of automated systems [1–6] envisage a certain composition and content of works performed at various stages of AS development (Table 1).

It is known in correspondence with provisions of guidelines that «process of AS development is a combination of ordered in time, interconnected, and integrated into stages works, fulfillment of which is necessary and sufficient for developing AS that meets the assigned requirements» [3, item 1.1]. It may be added that financing of AS development is usually performed in advance stage by stage. By the time, when acceptance tests are performed, as a rule up to 90 % of allocated resources, are used. At the same time at the stage of trial operation and acceptance tests shortcomings may be detected, which require for not just remedy of malfunctions and changes in the documentation, but also for specification of terms of reference.

However, such exceptional cases are caused by a certain reason. This is connected with the fact that during closing (acceptance) of a next in turn sub-stage of works associated with construction of AS specialists of the customer not always can estimate to full degree completeness and quality of the works carried out by the manufacturer. Criteria of estimation for them are usually results obtained after performance of welding — quality of a formed weld and presence of defects in it. Stage by stage fulfillment of the works causes sometimes situation when the most serious defects are detected at the end of the project implementation. The following trend is observed: the more complex is welding process the higher is the risk of force-majeur situation occurrence at the final stage of welding TPACS development.

In connection with mentioned above the need occurs to look for other organizational forms of project management when developing complex welding TPACS.

Specialists of the E.O. Paton Electric Welding Institute and Company «A.M. Makarov YuMZ» prepared proposals on using object-oriented design (OOD) when developing complex welding TPACS for missile carrier structures. As it is known, OOD has the following distinctive features [7, 8]:

- it is a return process (return design);
- continuous modeling of the result is implemented in it (availability of many AS versions);
- it uses evolution character of the design — from general requirements established for AS to more and more detailed ones;
- it requires for application of expressive (self-documenting) means of automated design of CAD type.

Difference of OOD from traditional approach to design is shown in a simplified form in Figure 1.

So, we may note that in the process of AS development using OOD a number of versions of software is developed. At the same time hardware (a complex of technical means) is improved and gradually approaches final version of AS.

Examples of efficient application of OOD in welding may be development of an automatic system for controlling TIG welding of vessels and pipes of big diameter [9] and complex-automated technologies of MIG and TIG welding [10].

Such developments have a number of distinctive features: increased complexity of automation task; wide use of computer technology (database control systems, CAD / CAM systems, etc.) both at the stage

**Table 1.** Typical composition of works when developing AS

| Stage | Sub-stage |
|---|---|
| Formation of requirements to AS | 1.1. Investigation of the object and substantiation of the AS development need<br>1.2. Formation of the user requirements to AS<br>1.3. Making out report on the fulfilled work and application on AS development (terms of reference) |
| Development of AS concept | 2.1. Study of the object<br>2.2. Performance of necessary scientific-research works (SRW)<br>2.3. Development of AS concept versions and selection of the option, which meets requirements of the user<br>2.4. Making out report on the fulfilled work |
| Terms of reference (TR) | 3.1. Development and approval of TR on AS |
| Conceptual design | 4.1. Development of preliminary design solutions on AS and its parts<br>4.2. Development of documentation |
| Detail design | 5.1. Development of design solutions on AS and its parts<br>5.2. Development of documentation<br>5.3. Development and making out documentation on supplying items for completion of AS and / or TR for their development<br>5.4. Development of tasks for design in contiguous parts of the project of the automation object |
| Operation documentation | 6.1. Development of operation documentation on the system and parts thereof<br>6.2. Development or adaptation of the programs |
| Commissioning | 7.1. Preparation of the automation object for AS commissioning<br>7.2. Training of the personnel<br>7.3. Completion of AS with supplied items (software and hardware, program-technical complexes, and information items)<br>7.4. Construction-assembling works<br>7.5. Starting-up and adjustment works<br>7.6. Preliminary tests<br>7.7. Trial operation<br>7.8. Acceptance tests |
| Support | 8.1. Fulfillment of works in correspondence with warranty obligations<br>8.2. Post-warranty servicing |

of AS modeling and at the stage of its development; application of mathematical modeling of welding and assembly processes (for example, «virtual factory» in [9]); continuous carrying out of scientific-research works for the purpose of increasing productivity of welding.

A formal sign of OOD is introduction of versification (or sequence) of the AS working prototype. For example, now we may speak about alpha- or beta-versions of AS (or first phase, second phase, etc.). For each version of the AS prototype tests may be carried out, and degree of its correspondence to TR may be estimated.

Approximate scheme of the welding TPACS development project using OOD is shown in Figure 2. New feature in this approach is that in case of detecting at one of intermediate sub-stages of a serious shortcoming in implementation of AS, development of the next version may be started from the stage of SRW fulfillment and correction of TR.

In the most general form development of AS, when using OOD, may include fulfillment in parallel of several works: for example, conceptual design, development of working documentation, and manufacturing of non-serial automation equipment components (AEC) may be carried out in parallel with SRW.

It is clear that with such organization of works greater efforts have to be applied to coordination of actions of the manufacturers than in case of the traditional approach. Availability of more specific criteria for estimating degree of the project termination is also needed. The latter may be achieved by «parameterization» when estimating readiness of the current AS version. Such parameters may be indices of quality and / or efficiency of welding TPACS functioning, for example, contours of automatic regulation (control of the torch position relative a butt, a weld width, the depth of penetration, position of a tip with a filler

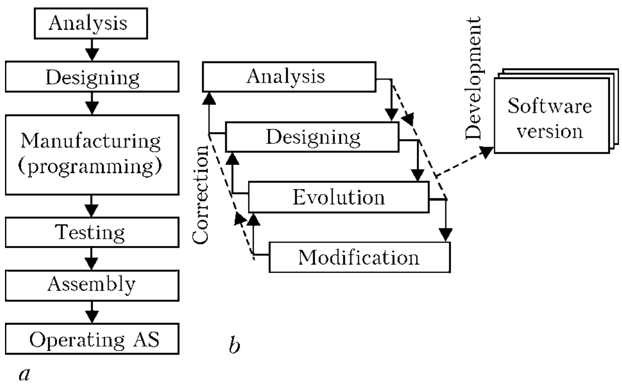


**Figure 1.** Comparison of approaches to AS design — traditional (*a*) and OOD (*b*)

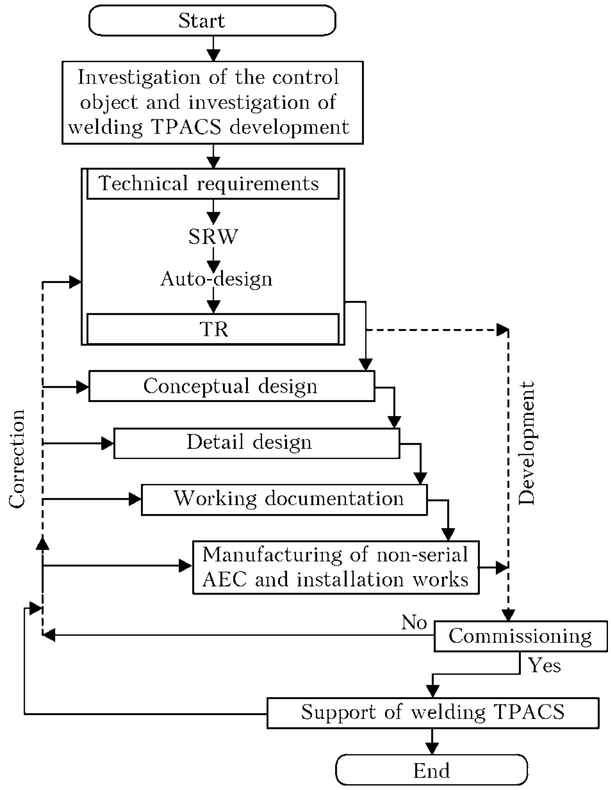


**Figure 2.** Approximate scheme of welding TPACS project implementation using OOD

wire, etc.), functioning modes (manual, automatic, test, etc.), a welding cyclogram («firing» of the arc, welding, welding up of a crater), technological order of welding (first and second runs, weld quality control, etc.), completeness of AEC, etc. System of such parameters may be developed with attraction of the customer specialists, and due to this it will be quite understandable for them, because it will relate to the subject area of this kind of welding. After the system of AS parameters is coordinated, one can rather simply estimate degree of the AS readiness in percent in regard to each quality index.

It is evident that such approach allows a customer adequate estimating at any stage of fulfillment of the works consistency of the project and operative controlling progress in fulfillment of the project in regard to various parameters. One more advantage of such approach is the fact that due to periodic performance of tests of intermediate versions of AS prototypes one can make decision on beginning of the trial operation of welding TPACS in reduced mode (for example, in manual mode with operating contour of automatic directing a torch at the butt).

Approximate structure of the calendar plan of works in fulfillment of such task is given in Table 2.

Singularity of the calendar plan form consists in the fact that it has no names of certain stages and sub-stages of the AS development works, which are stipulated in item 2.1 [3]. In reality they exist, but in implicit form. Fulfillment of various stages of the work is connected with achievement of certain value levels (degree of readiness) of a group of functioning indices.

Analysis of the condition of guidelines of inter-government significance (state standards) showed that proposed project implementation scheme does not contradict their concept. For example, the same document [3] reads that «depending upon specificity of the AS being developed and conditions of their development it is allowed to fulfill certain stages of works before termination of the preceding stages, parallel in time fulfillment of stages of the work, and inclusion of new stages of work» (item 2.2).

In connection with mentioned above we may consider that, firstly, in our case certain stages of works are fulfilled in parallel and, secondly, several stages of works are added (see items 2--4 in Table 2).

Analysis of national guidelines on development of AS does not allow making unambiguous conclusion concerning possibility of OOD application when developing welding TPACS.

One can note a system of national standards [11--15], which concern development and putting of items on production. It follows from item 5.1.2 of [12] that AS relate to the field of action of these standards. Analysis of the content of mentioned documents showed that application of OOD in development of AS is difficult, because availability of separate normative documents on fulfillment of SRW and research and development works makes it impossible to implement principle of the return design. However, in item 5.3.5 of [12] it is noted that development of automated control systems is performed in accordance with a complex of standards and guidelines on AS. The latter

**Table 2.** Approximate calendar plan of works developed using OOD

| Stages | Degree of readiness according to functioning indices, % | | | | | |
|---|---|---|---|---|---|---|
| | A | B | C | D | ... | Z |
| Analysis of technical requirements and results of welding technology investigation, development of TR | 5 | 0 | 5 | 0 | ... | 5 |
| Development of alpha-version of welding TPACS | 25 | 10 | 25 | 0 | ... | 10 |
| Development of beta-version of welding TPACS | 50 | 25 | 50 | 75 | ... | 75 |
| Development of final version of welding TPACS | 95 | 95 | 100 | 100 | ... | 95 |
| Performance of acceptance tests | 100 | 95 | 100 | 100 | ... | 100 |
| Support and removal of remarks | 100 | 100 | 100 | 100 | ... | 100 |

may be interpreted as limitation of action of this system of national standards and expansion of limits of using inter-state standards for development of AS.

It should be noted in conclusion that application of OOD in development of AS ensures the following additional advantages: the AS being developed is characterized by more open properties (it corresponds to requirements [2, 5]) and as a result of the project implementation cognitive capacity of specialists-developers is activated.

So, one may draw conclusion that there are objective premises for using OOD in development of complex welding TPACS.


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3. *GOST 34.601–90*: Information technology. Complex of standards for automated systems. Automated systems. Stages of development. Introd. 01.01.92.
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# ECONOMIC EFFECTIVENESS OF RAILWAY WHEEL PROFILE RESTORATION


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Quantitative evaluation of the influence of various factors on the effectiveness of wheel profile restoration by cladding, and, in particular, after annealing of the rolling surface hardened in operation. Effectiveness of wheel profile restoration is increased by applying annealing of the rolling surface, multi-electrode cladding and heat treatment of the clad and turned wheels.



*Keywords: effectiveness, restoration, railway wheels, rolling surface defects, HF current annealing, multi-electrode cladding*


Annually 3–3.5 million of railway wheels are manufactured in the world according to VNIIKI (Moscow, Russia) data, of which 40–50 % are manufactured and used in CIS countries, which is indicative of their very low actual residual life (although the design life of railway wheels, manufactured in CIS, is 12 years). The service life of the wheels is mainly determined by turnings, when a much larger amount of metal is cut away, than as a result of operation. Carriage wheels that have a worn out rolling surface after service (Figure 1), are restored to initial profile (Figure 2, pos. *1*) by turning on wheel lathe by a template [1] and in such a way the rim thickness is decreased.

At the beginning of 1990s because of intensive wear of ridges and corresponding decrease of rim thickness when turning of wheel profile for restoration, their residual life was reduced to 4–5 years [2]. In 1994 the average life of carriage wheel pairs (between turnings) that were sent in for maintenance because of ridge wear was 10.6 months. 4.6 % of them had the service life of up to 3 months, 16.5 % — from 3 to 6 and 17.8 % — from 6 to 9 months [3]. As the result of these measures directed to improve the conditions of work (toughening of norms for maintenance of the track and operation of the rolling stock, improvement of repair technology and maintenance of carriages, rails lubrication for curvilinear sections; use of rolled shapes optimum for the rail profiles [2]), wear of wheel ridges (rim thinning is determined at the distance of 70 mm from inner side of the wheel) and rail heads decreased 2 times. Cladding of ridges of railway wheels before wheel turning (Figure 2, pos. *2*) allowed

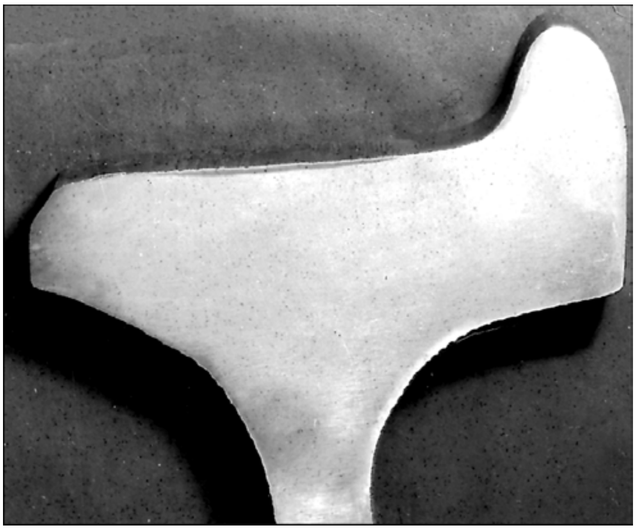

**Figure 1.** Macrosection of wheel fragment with worn rolling surface

decreasing the thickness of cut away rim layer [4, 5]. However, no adequate increase of wheel service life was observed. The average service life increased up to 6–8 years [2].

The efficiency of ridge restoration by cladding influences not only the price level of new wheels and the expenditures on their restoration, but also the level of ridge wear, depth of defect location in the rim, that determines the decrease of rim thickness as a result of turning operations. At present time ridge cladding is done without preliminary turning of defects on the rolling surface, that sharply decreases the cladding efficiency. Despite the decrease of defect number on the rolling surface the thickness of removed layer in turning of wheels with defects along the rolling circle did not decrease, more over, problems appeared in turning wheels of a higher hardness.

Today in Russia, Kazakhstan and Ukraine simultaneously with production of higher hardness wheels [6, 7], the idea of manufacturing cast wheels is being actively popularized [8], for example, by the technology of Griffin Wheels company, USA [9]. At first glance, the advantage of such a technology is connected with a considerable saving of means. At increase of the run up to 1 mln km of American railways, it is cheaper to remove the wheels from operation than to carry out their turning. It, however, envisages a multistep, expensive system of monitoring in manufacturing and operation, the need for a frequent exchange of expensive moulds (one mould for 1000 wheels). The most important is the fact that cast wheels of Griffin Wheel company are more brittle, because they are manufactured from steel with a high content of carbon (0.67–0.77 %), that does not exclude the possibility of appearance of surface defects of brake origin or their spreading into the wheel in operation in railways of the post-Soviet space.

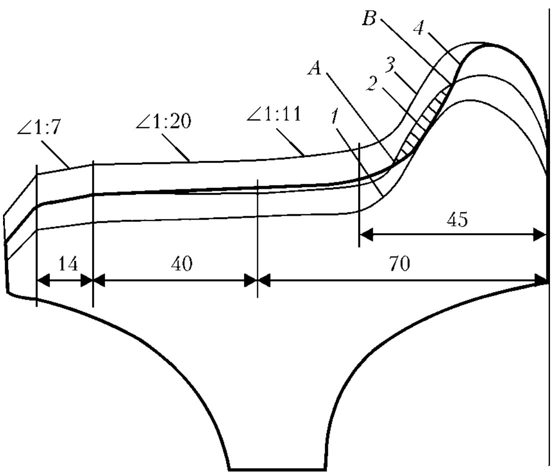


**Figure 2.** Diagram of wheel profile restoration: *1, 2* — lines of turning after ridge cladding (*A* — start; *B* — end of cladding); *3* — line of turning after wheel restoration by ridge and rim cladding; *4* — line of worn wheel surface

Technical solution of Valdunes company, France, that is the world leader in developing of railway wheels manufacturing technologies, is connected with application of medium and low-carbon wheel steels, with carbon content of 0.22 % and alloyed by chromium, manganese and silicon [10]. Such chemical composition allows limiting martensite formation. At the same time at XXIII International Congress on Wheel Pairs, held in Rome at the end of 2001, it was noted that economically it is more profitable to use wheels with a low-carbon bainite steel coating [11]. Despite formation of flats, no cavities or cracks develop in this steel, that gives the possibility to decrease the thickness of the cut off layer at turning the wheels and to prolong the term of operation. The present period in Russia as well as in Ukraine is characterized by a considerable decrease of work volumes on ridge restoration and wheel cladding. Moreover, the technologies of profile restoration of railway wheels by cladding of the entire rolling surface (Figure 2, pos. *3*) using low-carbon wire [5, 12, 13] are not introduced in repair of rolling stock, that operates in the main railway lines of CIS countries.

The goal of this study is the quantitative evaluation of factors, that influence the efficiency of wheel profile restoration by cladding, in particular, after annealing of the rolling surface, strengthened in the process of operation [14–16].

At restoration of 1 mm of worn ridge of railway wheels to a standard profile with 33 mm ridge width (at the distance of 18 mm from ridge top), it is necessary to decrease the rim thickness by 2 mm. Data on distribution of 79235 wheels that were delivered in 2003–2004 to the sections for ridge cladding at Viltrans Ltd. of Ukrzaliznytsya Company Carriage Repair Plants, by ridge thickness, as well as rim thinning of these wheels at their restoration by turning are given in the Table (Figure 2, pos. *1*).

Taking into consideration such a distribution of parameters of the wheels to be repaired, ridge cladding before turning allows preserving 13.3 ± 55 % of wheel rim diameter on average, that should increase their residual life, accordingly. However, because of the need for defect turning before ridge cladding, the efficiency of cladding decreases substantially. In practice all the wheels that are sent in for repair, have

| | | | |
|---|---|---|---|
| Thickness of wheel ridge sent in for repair, mm | 22–24 | 25–27 | 28–30 |
| Necessary decrease of wheel rim thickness at restoration of profile by turning, mm | 22–18 | 16–12 | 6–5 |
| Distribution of wheels, sent it for ridge cladding, depending on ridge thickness, % | 15 ± 5 | 68 ± 3 | 17 ± 12 |

2–3 flats of the depth of less than 2 mm. Statistics shows that about 25 % of the wheels have flats with metal hardness on railway wheel rolling surface of not less than *HRC* 40–50. At restoration by cladding of wheel ridge with rim defects, the efficiency of restoration consists in preserving 5–6 mm of the rim thickness, that makes 38–46 % of average preserved rim thickness. If we assume that wheel turning is performed once per year, then average life of wheels with defects will be 6.5 years even after applying cladding of worn ridges.

Let us determine the efficiency of restoration by cladding of the wheel profile with worn rolling surface for the case of a real wheel that was sent in for repair (see Figure 2). Rim thickness that was measured in the rolling spot is 27 mm. At the same time the thickness of worn ridge at 18 mm height is 25 mm (wear is 8 mm) and the required decrease of ridge rim thickness at its restoration by turning in the point of roll measurement is 13 mm (determined by line *1*), whereas rim thickness after turning is 14 mm. After ridge cladding it is necessary to perform turning of worn wheel by its profile (line *2*) from bevel to ridge side on the depth of 4 mm in the point of roll measuring (including turning of the clad ridge on the line between points *A* and *B*). In this case rim thickness will be 22 mm. Efficiency of cladding the given ridge after rim turning (without taking into account the rim thinning at defect turning) consists in preserving 7–8 mm of rim thickness (53–61 % of average value of preserved rim thickness equal to 13.3 ± 5 %).

The wheel shown in Figure 1, besides the worn rim and ridge, has a defect on the rim (a flat 1.5–2.0 mm deep with hardness of about *HRC* 50). After ridge cladding and final turning of the wheel by 1 mm (see Figure 2) the rim thickness is 17–18 mm. The wheel after restoration is not ready for operation, because its rim thickness is less than 22 mm as the result of turning (the efficiency of restoration is equal to zero, as wheels with rim thickness of not less than 22 mm are allowed for operation).

It is known that annealing of wheel rolling surface allows lowering metal hardness minimum 2.5 times, that essentially decreases the chip thickness at their turning [17, 18]. This feature was used by us at wheels restoration by cladding of the rolling surface. Tensile stresses that appear after annealing by HF current, are favorable for revealing and rejecting «weak» wheels with microdefects. Since 1997 as a result of rolling surface annealing, turning to the depth of detected defects, rejection of wheels with cracks on the surface (0.2 % of wheels sent in for repair), with further cladding of worn ridges in wheel repair plants of Ukraine has allowed restoring about 270 thousand of wheels for further operation [15]. Annealing of the hardened metal of wheel rolling surface before cladding is favourable for refining of metal grains and decreasing the probability of cold crack formation in the zone, adjacent to the fusion line, that allows cladding to be performed without defect turning on the rolling surface, that is without loss of wheel life [16, 17]. After annealing of rolling surface and turning of defects on the rim before ridge cladding, the total volume of expenditures (welding consumables, time, power) required for cladding, decrease by 30 %. It is possible to achieve a considerable saving in the process of wheels heating (before cladding) due to using one installation for annealing (up to 60 wheels per 8-hours shift). Introduction of wheel ridge and rim thickness control after defect turning allows assessing the expenditures and efficiency of ridges cladding for each wheel.

The efficiency of wheel profile restoration by worn ridge cladding is determined by correlation of the preserved rim price and value of expenditures for wheel profile restoration (rolling surface annealing and cladding of worn areas, turning, etc.), and it is calculated in accordance with the formula below:

$$E = [(H_{rim} - H_{d.t})\ C_w - P_{cl.an}] / P_{cl.an}, \qquad (1)$$

where $H_{rim} = 2(33 - H_{rid})$ is the rim decrease at turning the wheel without cladding, mm; $H_{rid}$ is the thickness of worn ridge at the height of 18 mm from the top, mm; $H_{d.t}$ is the rim decrease at defect turning, mm; $C_w$ is the reduced cost of 1 mm of the rim, UAH / mm; $P_{cl.an}$ is the price of cladding and annealing ($P_{cl.an} = P_{cl} + P_{an}$), UAH.

Restoration of wheel profile of worn out ridge by cladding is profitable if condition $[(H_{rim} - H_{d.t})C_w - P_{cl.an}] \geq 0$ and $E \geq 1$ is fulfilled.

Dependences of the efficiency of restoration for wheels with different wear of the ridge at ridge thickness of 23, 26, 29 mm on the rim decrease at turning of worn and defective areas are given in Figure 3.

Basic data for calculation of the efficiency of wheel restoration by ridges cladding at constant values of $C_w$ (USD 1 = UAH 5.05) are the following: the price of a new wheel by 01.09.2005 without VAT — UAH 2,650; useful life of the rim — 52 mm; $C_w$ = = UAH 51; $P_{cl}$ = UAH 72.67; $P_{an}$ = UAH 16.25; $P_{cl.an}$ = UAH 88.92.

Cladding of wheel ridges 29 mm thick is not profitable, if thickness of the rim decreases as a result of turning by more than 4 mm, for 26 mm thick by more than 10 mm and so on.

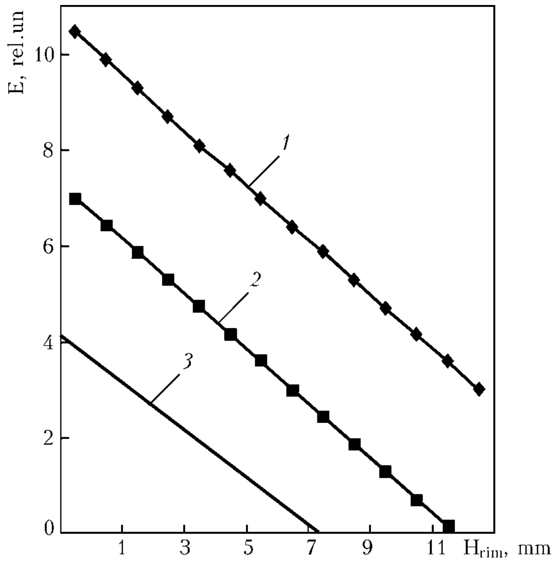


**Figure 3.** Dependence of the efficiency of ridge wheels restoration by cladding on rim thinning at turning of flats and defects: *1* — $H_{rid}$ = 23; *2* — 26; *3* — 29 mm

At restoration of wheel profile by turning by line *2* (see Figure 2) with subsequent cladding of the worn rim and ridge, the ridge thickness decreases by 1–2 mm and does not depend on wheel rim thickness decrease at turning. If the increase of wear resistance of the clad layer is ignored, it is possible to define an inverse problem (1) for determination of the mean price of restored wheels, that are sent in for repair with ridge thickness of 23, 26, 29 mm and to derive the corresponding dependences on the restoration efficiency at constant values of a new wheel cost:

$$P_{rest} = [(H_{rim} - H_{f.t})C_w] / (E + 1), \qquad (2)$$

where $P_{rest}$ is the price of wheel profile restoration by cladding of the worn ridge and rim; $H_{f.t}$ is the decrease of rim thickness at final turning of the clad wheel by the profile ($H_{f.t} \le 2$ mm).

For example, if $C_w$ = UAH 51, wheels restoration is efficient, i.e. $E \ge 1$ at $P_{rest} \le$ UAH 302 (about 10 % of new wheel price). In the example, given in Figure 2, after restoration of the wheel with rim thickness less than 27 mm, which is to be pressed out for scrap (see Figure 1), rim thickness is more than 32 mm, that enables the wheel to run minimum till the first turning.

In study [16] it is shown, that high efficiency of restoration of the wheel profile after cladding of the worn ridge and rim can be achieved by applying:

• annealing of wheel rolling surface in HF current installation with 10 kHz frequency to the depth of about 2.5 mm, that provides the possibility of increasing divergent stresses, detecting defects and rejecting the wheels, lowering the probability of cracks initiation after cladding, reduction of thickness of the cladding layer, required for restoration;

• grinding the locations with deep cracks to the depth of more than 3 mm instead of turning by rolling circle;

• multi-electrode one or two layer cladding [19] on wheel rim [20] by ten electrodes into one weld pool with alloyed wire, that ensures a high efficiency and quality of cladding metal (not less 6--8 grain point in the coarse-grained section of HAZ metal);

• cladding of worn ridges;

• heat treatment of clad and turbine wheels in HF current installation, providing the level of their divergence after cutting of not more than ± 3.5 mm.

Cladding of low-carbon metal with higher wear-resistant characteristics allows decreasing defect propagation on the rolling surface in service, reducing the thickness of the cut layer during wheels turning and increasing their residual life. Restoration of thin rimmed wheels with rim thickness less than 30–35 mm, that were earlier scrapped, allows increasing their service life not less than 2 times.


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# EFFICIENT ORGANISATIONAL STRUCTURES FOR TRAINING WELDING SPECIALISTS FOR SHIP BUILDING

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The article describes positive experience of establishment and efficient forms of activity of a regional centre for training welding fabrication specialists for ship building in Ukraine.

*Keywords: welding fabrication, ship building, training of staff, organisational structures, efficiency*

Ukraine takes an active part in the process of integration into the European Community, including on the basis of international experience and trade practice inside the country. This applies to the full extent to the field of training welding fabrication specialists [1].

For example, the Admiral Makarov National Ship Building University was approved in 1996 by the joint order of the Ministry of Education and Science and Ministry of Industry of Ukraine to be the leading institution to deal with training of all levels of welding fabrication specialists for ship building.

In this connection, in 1996 the Welding Fabrication Chair of the University, together with shipyards of the South of Ukraine, founded Association «Ship Building Welding Training Centre». Co-founders of the Centre are Open Joint Stock Company «Damen Shipyards «Ocean», Research-and-Production Gas Turbine Construction Company «Zarya-Mashproekt», Kherson Ship Building Factory, «61 Communards» Factory, Chernomorsky Ship Building Factory and Admiral Makarov National Ship Building University.

The Association is an independent legal entity, the activity of which is based on the self-financing and self-government principles. The highest authority of the Centre is the Board of Trustees. To widen practical training of engineers and adapt them to the fabrication practice, the Chair established two subsidiaries at the leading enterprises of the city — Open Joint Stock Company «Damen Shipyards «Ocean» and Research-and-Production Gas Turbine Construction Company «Zarya-Mashproekt». Directors of the Chair subsidiaries are the chief welders of the enterprises Yu.V. Solonichenko and Yu.V. Butenko. These enterprises are also the bases for all types of practical studies of the students.

Establishment of the Association allowed the Chair to substantially expand the sphere of its activity and co-operation with industrial enterprises. In 1997, under the International Program «Transform», the Association, together with the German Research and Training Centre (SLV) of Meklenburg Vorpommern, carried out re-training, in conformity with the European Standards, of a group of 24 leading welding fabrication specialists from the ship and machine building enterprises of the South of Ukraine by granting them the «European Welding Engineer» certificate. In 2000, under the agreement on co-operation with SLV, chief welders of the ship building factories were trained according to the program «International Welding Engineer». This program allowed lecturers of the Chair to improve their qualification and obtain the corresponding certificates [2].

The Association provides training and certification of welders according to DNAOP 0.00-1.16--96 (Resolution # 019.05.48.80.42.0) and rules of the Russian Register of Shipping (RRS) (enterprise certificate # 01.030.160). Training and certification allow performance of the work associated with fabrication of different types of structures, including manufacture, assembly, reconstruction and repair of facilities and equipment in accordance with standards in force in Ukraine (DNAOP, SNiP, DBN, DSTU), as well as facilities supervised by RRS. For these purposes, the Association has a permanent certification commission chaired by Yu.V. Butenko (chief welder of the Company «Zarya-Mashproekt»), Zh.G. Goloborodko (chief welder of the Kherson Ship Building Factory), Yu.M. Konashchuk (chief welder of CNGS Engineering, Simferopol), and A.M. Kostin (associate professor of the Admiral Makarov National Ship Building University). Over 100 welders were trained and certified in 2004, which made the organisational work of chief welder's departments of the southern region enterprises much easier. The Association also provides training of the first and second level specialists in non-destructive testing of welds.

Experience of functioning of the Association shows that it is beneficial for enterprises to certify welders on a unified base, having a substantial materials and scientific potential. The Chair assists much in handling important production problems and helps to involve funds for practical training of students, repair and upgrading of equipment, classes, lecture rooms, etc. For example, since 1996 all consumables (metal, electrodes, working cloths, masks, etc.) needed for training of the students to master a skill of welding operator have been bought at the expense of the Association. Part of the students passes qualification exams and receives basic welder certificates, which allows them to work during their practical studies by profession.

Under the program on training and certification of welders, the Chair holds workshops and consultations on welding consumables by involving leading national and foreign companies, which allows specialists of enterprises to exchange their working experience and obtain new knowledge.

Working as independent experts, specialists of the Chair systematically assess the quality of fabrication of individual units and facilities, technologies (DSTU 3951--2000), and assembly-welding production as a whole. This makes it possible to conduct collaborative research and apply advanced technical solutions in production. The most demonstrative example in this respect is the joint development by the University and Research-and-Production Company «UkrTermMash» of a new generation of thermal cutting machines of the «Kristal--TM MPlK» type with a computerised control system, which are applied at more than 20 enterprises of Ukraine and other CIS countries [3].

The Association assumed the functions of a regional co-ordinator and consolidated efforts of enterprises in the field of training of all levels of staff for welding fabrication. This widens the sphere of activity of the Chair and strengthens contacts of the team of lecturers with successfully operating enterprises.

The fruitful co-operation has been established lately with our constant partners, such as SELMA, Kakhovka Plant for Electric Welding Equipment, CNGS Engineering, Damen Shipyards «Ocean», «Zarya-Mashproekt», Kherson Ship Building Factory, etc. These enterprises do not only invite our graduates to work, but also delegate their workers to have training by correspondence. Diplomas prepared in the last three years by orders of enterprises constitute 53.6–78.9 %. Representatives of the enterprises attend defence of the diplomas and participate in selection of the best of them, and choose the best graduates on a competition basis. Demand for welding fabrication specialists has substantially grown lately. Now the demand is in excess of supply, and all students find employment in their specialities.

The Chair maintains permanent contacts with graduates working at enterprises of Nikolaev, Simferopol, Kerch, Kherson, Odessa and many other cities, invites young specialists to participate in conferences, involves them in co-operation to address topical problems, and invites for post-graduate courses. Now there is a real possibility to combine a joint research activity and professional growth of enterprise associates.

Kakhovka Plant for Electric Welding Equipment (Kakhovka), SELMA (Simferopol), «Zarya-Mashproekt» and Damen Shipyards «Ocean» provide a substantial support to the Association. In the last years the Chair has received new equipment, materials and sponsor assistance for an amount of more than 70,000 UAH.

As shown by experience, the integrated approach to the problem of staff training, based on close co-operation with enterprises of different forms of ownership, establishment of subsidiaries of the Chair at leading industrial enterprises, and arrangement of the regional staff training centre, provides widening of the sphere of activity of the Chair and upgrading of material and technical base for quality training of all levels of welding fabrication specialists to meet the needs of the industry.

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# FLUX-CORED WIRE FOR SURFACING OF MARAGING STEEL LAYER

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Described is a new flux-cored wire PP-AN204 that provides deposited metal of the type of maraging steel with the Fe–Ni–Mn–Si–Mo alloying system. Hardness of the as-deposited metal is *HRC* 20–30, which allows its easy cutting, while after tempering its hardness grows to *HRC* 50. After tempering, the deposited metal is characterised by high hot hardness, thermal stability and wear resistance in friction of metal on metal at increased temperatures. Therefore, it can be recommended for hardening of working surfaces of complex-configuration die tools.

*Keywords: arc surfacing, deposited metal, maraging steel, flux-cored wire, hot hardness, wear resistance, heat resistance*

Steels and alloys with an effect of secondary hardening resulted from their tempering within certain temperature-time range are rather broadly applied in different branches of industry [1, 2]. It was proposed to use such materials (particularly maraging steels) also in the surfacing production mainly for restoration and hardening of tools for hot deformation of metals [3–5]. However, due to the fact that hardness increase in most of these materials was only *HRC* 10–15 they did not find a wide application in the industry. Besides, some of them initially had high hardness above *HRC* 40 or high cost because they were alloyed with expensive elements (tungsten, cobalt, molybdenum) with their total content being 20 % and higher.

In is shown in work [6] that save-alloyed deposited metal of Fe–Ni–Mn–Si–Mo system has a considerable effect of secondary hardening (hardness after surfacing is *HRC* 29–30, after tempering is *HRC* 50–52), so this material is promising for restoration and hardening of tools for metal hot deformation. The results of study of the multi-layer surfacing effect on hardness distribution on the fusion boundary of neighboring beads and layers, as well as wear resistance, thermal resistance and hot hardness of the deposited metal of this alloying system are presented in this work.

During multi-layer surfacing of maraging steels a heterogeneity in hardness distribution in the HAZ metal near the boundary of fusion of neighboring beads and layers as a result of heating up to temperatures 480–500 °C providing secondary hardening is observed, which deteriorates operational properties of deposited metal and complicates its machining.

To study these phenomena a multi-layer surfacing of St3 steel plates 20 × 50 × 200 mm in size was carried out using experimental self-shielding flux-cored wire, which produces deposited metal of maraging steel type of the preset alloying system. Surfacing with wire 2.0 mm in diameter is carried out under the following conditions: current is 300 A, voltage is 24–26 V, surfacing speed is 20 m / h, surfacing step is 9–11 mm (overlapping of about 30 %) and 7–8 mm (overlapping of about 60 %). Templates for producing microsections for investigations were cut out from the deposited plates across the deposited beads. Microsections were used to measure microhardness (load $P$ = 1 kg, step is 0.5 mm) by depth of multi-layer deposited metal (Figure 1) and in the fourth deposited layer across deposited beads (Figure 2) before and after aging.

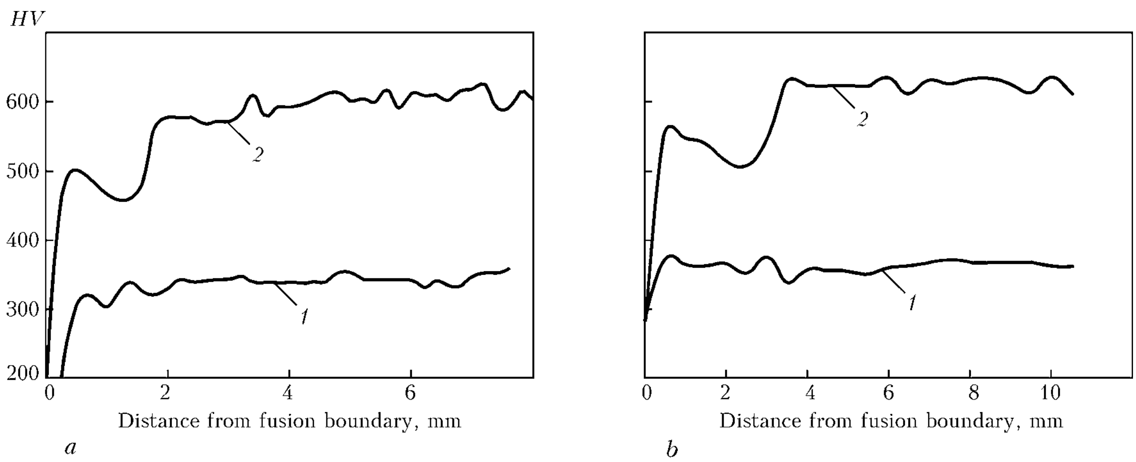


**Figure 1.** Distribution of microhardness by depth of maraging deposited metal layer: *a* — overlapping 30 %; *b* — 60 %; *1, 2* — before and after aging, respectively

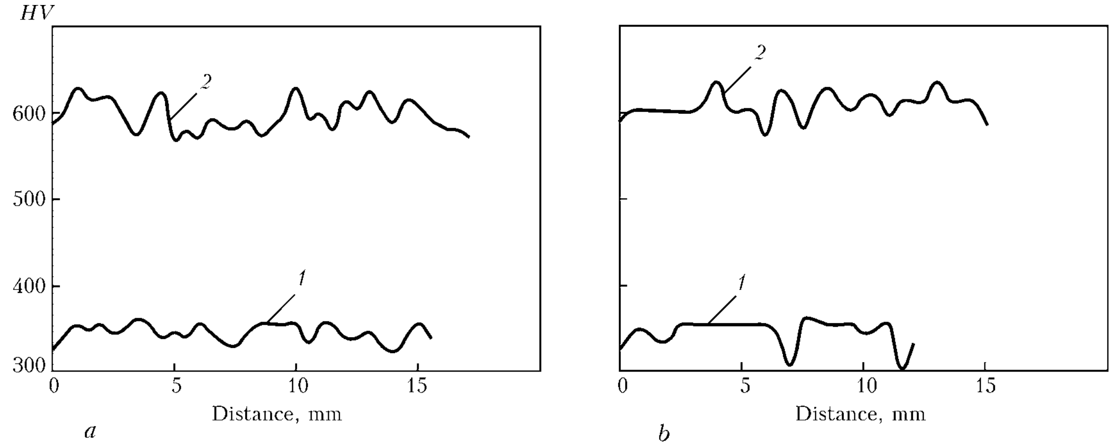


**Figure 2.** Distribution of microhardness crosswise in the fourth layer of the deposited maraging metal: *a* — overlapping 30 %; *b* — 60 %; *1, 2* — before and after aging, respectively

Overlapping of deposited beads did not produce an essential effect on distribution of microhardness by the depth of the deposited metal. Irrespective of overlapping directly after surfacing microhardness by the depth of the deposited metal varies within *HV* 10–20 and achieves values typical for deposited metal of the preset composition (see Figure 1, curves *1*) already in the third layer. Spread of deposited metal hardness after aging remains within the same ranges — *HV* 10–20 (see Figure 1, curves *2*).

Surfacing step has an inconsiderable effect on the hardness distribution across the deposited beads, though spread of hardness on the boundary of neighboring beads is somewhat higher (see Figure 2, curves *1, 2*). Nevertheless, no considerable increase of hardness is observed here (due to relatively short time period of the temperature effect) and its variations in this case are also within the acceptable ranges.

Thermal resistance tests were carried out on the unit for comprehensive evaluation of the deposited metal properties. The unit was developed by the E.O. Paton Electric Welding Institute [7]. Specimens $40 \times 40 \times 40$ mm in size were cut out from blanks surfaced with experimental flux-cored wire. Thermal resistance was estimated by the number of heating-cooling cycles of polished surface of the deposited specimen until net-shaped roll marks visible with the naked eye appeared. The specimen was heated by a gas torch up to the temperature 650–700 °C with subsequent water cooling down to 60–70 °C. On the average the results of the tests of 3 specimens show that net-shaped roll marks appear on the surface of the maraging deposited metal specimen after 40 heating-cooling cycles (Figure 3). As it is seen from the presented data, the experimental maraging deposited metal exceeds by this index the known types of deposited metals (35V9Kh3SF, 150KhNM and others), which are successfully applied for surfacing tool for hot deformation of metals.

Tests on wear resistance under friction of metal against metal at increased temperatures (scheme of tests is shaft–plane) were carried out on the same unit [7] in the following conditions: specific pressure in the contact site is 80 MPa; friction speed is 11–12 m / min; temperature of ring-counterbody is 800 ± 30 °C; temperature on the surface of the tested specimen in the contact zone is 500–550 °C; time of testing is 1 h. Flame of gas torch was used as a heating source of the abrasive ring. Dimensions of the ring-counterbody produced of hardened steel 45 are as follows: diameter is 110 mm, width is 30 mm, thickness is 20 mm; dimensions of the specimen are $10 \times 20 \times 40$ mm. Specimens of hardened steel 45 were also tested as standard for comparison. Wear resistance was estimated by the loss of weight $\Delta G$ of the tested specimen and abrasive ring before and after wear. As wear resistance experiments show the ex-

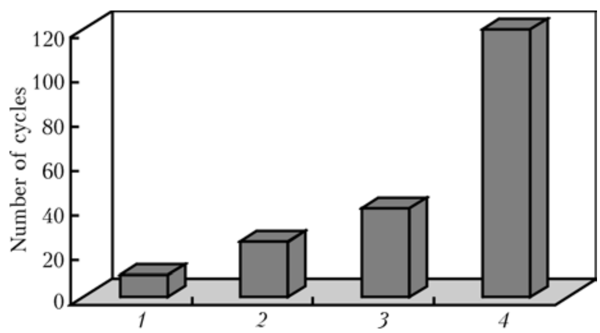


**Figure 3.** Thermal resistance of deposited metal: *1* — 150KhNM; *2* — 35V9Kh3SF; *3* — experimental maraging steel; *4* — 25Kh5FMS

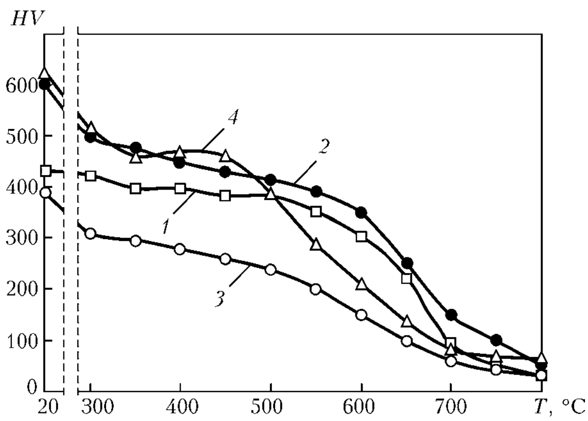


**Figure 4.** Hot hardness of deposited metal: *1* — 25Kh5FMS; *2* — 35V9Kh3SF; *3* — 150KhNM; *4* — experimental maraging steel

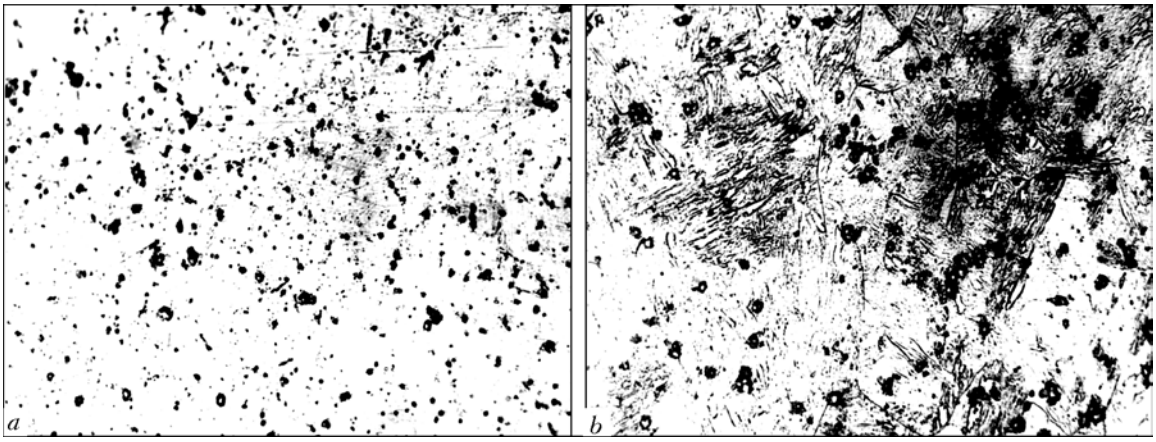

**Figure 5.** Microstructure of experimental maraging deposited metal before (*a*) and after (*b*) aging. Electrolytic etching in 20 % chromium acid (×400)

perimental maraging deposited metal 3.5–4 times exceeds the hardened steel 45.

Hot hardness of the experimental deposited metal was studied in comparison with the deposited metals of the known tool steels (Figure 4). Specimens were vacuum-heated in the special inductor, hardness was measured with the load of 1 kg and holding of 60 s. It is seen from the presented data that hot hardness of maraging deposited metal is on the same level with hot hardness of chromium-molybdenum and chromium-tungsten die steels surfaced with corresponding flux-cored wires.

Metallographic studies of the experimental maraging deposited metal show that its structure prior to aging is composed of ferrite, small quantities of martensite and titanium nitrides (Figure 5, *a*). After aging a share of martensite considerably increases, sites of fine-needled and lath martentise occur, the content of nitride increases (Figure 5, *b*). Microhardness of matrix before aging is *HV* 283, after aging is *HV* 489; microhardness of nitrides is *HV* 1206. X-ray structural analysis carried out on the unit «Dron 3» confirms that secondary hardening of experimental deposited metal is attributed to formation of considerable quantities of martensite.

Self-shielding flux-cored wire conditionally called PP-AN204 was developed for surfacing of a layer of maraging steel of Fe–Ni–Mn–Si–Mo alloying system. Welding technological properties of new wire were studied. Gas-slag-forming system $CaF_2$ + $TiO_2$ + $CaCO_2$ used in the flux-cored wire provides high arcing stability and minimal spattering. Splashing coefficient was within 12–14 %. Wire consumption per 1 kg of deposited metal was 1.25–1.27 kg. With wire diameter 2.0 mm a deposition coefficient was 15–17 g / (A·h). Deposited metal with a minimal quantity of slag inclusions has no pores and cracks.

In conclusion it is worth noting that the developed shelf-shielding flux-cored wire PP-AN204 allows producing a deposited metal of save-alloyed maraging steel type. The deposited metal of this type has high wear resistance (in conditions of friction of metal against metal at increased temperatures), thermal resistance and hot hardness. As a result of studies flux-cored wire PP-AN204 may be recommended for restoration and hardening of tools for hot deformation of metals, first of all, tools of complex shape requiring a lot of efforts for machining.


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# MODERN MECHANISMS OF ELECTRODE WIRE FEED IN MACHINES FOR MECHANIZED WELDING, SURFACING AND CUTTING

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Considered are current developments in the field of electrode wire feed mechanisms for welding, surfacing and cutting of steels and aluminium alloys. Classification of the main types of feed mechanisms is given. Of particular interest for welding specialists are the mechanisms that feed electrode wire with controlled non-stationary character of movement (intermittent, modulated or pulsed feed). Advantages and drawbacks of some ingenious designs of the pulsed mechanisms are considered.

*Keywords: mechanized arc welding equipment, electrode wire, feed mechanism, classification, feed methods, non-stationary process control*

Well-substantiated selection of equipment (semi-automatic machines for implementation of mechanized arc processes) is the leading link of welding production organization in fabrication of welded metal structures, repair of machines and mechanisms to extend the operating life, and lately also in cutting, particularly, if there is the objective of minimizing the energy- and resource consumption. The developers, designers and manufacturers of the mechanized arc welding equipment are continuously faced with the need to select the electrode wire feed mechanism.

This paper presents the modern developments of the mechanisms of electrode wire feed in welding, surfacing and cutting of steels and aluminium alloys.

Electrode wire feed mechanism in the modern design of a semi-automatic welding machine can have several main functions, namely wire feeding; technological process control; improvement of semi-automatic machine design.

Classification of the mechanisms of electrode wire feed in semi-automatic machines of different degrees of complexity designed for operation under different conditions and solving diverse technological problems, is shown as a schematic in Figure 1. In this classification the mechanisms for an intermittent and pulsed feed of electrode wire are regarded to be self-contained; they are considered on a par with the main type of mechanisms. Number of developments of the mechanisms imparting a non-stationary motion to the wire, is rising continuously. They are improved in engineering terms, and enable solving the urgent problems of welding fabrication. However, the gear unit with the roller pairs for a smooth feed of the wire remains to be the main one for semi-automatic machines of any type and purpose.

Systems of electrode wire feed with two pairs of feed and pressure rollers have become widely accepted

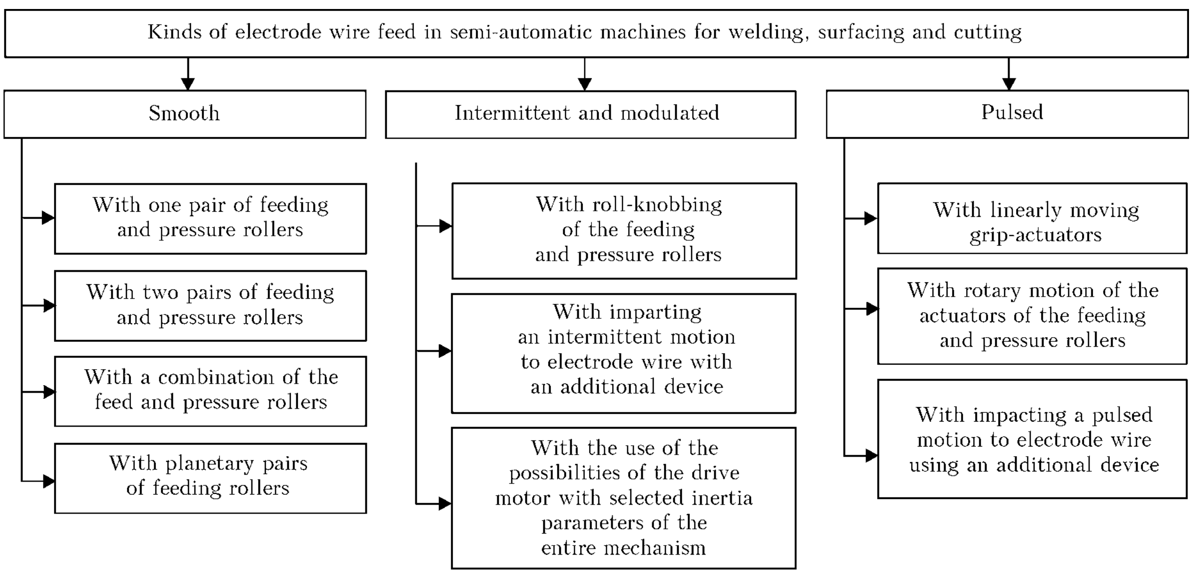


**Figure 1.** Classification of the main types of electrode wire feed mechanisms

lately. This is confirmed by the fact of manufacture of such mechanized arc welding equipment all over the world, where mechanisms with two pairs of rollers are used as the feed mechanism practically in all the designs of the semi-automatic machines, this being a kind of a fashionable solution, which is, however, not always justified. The functioning of such a mechanism can be made more complicated by the fact that the feed rollers have different angular speeds of rotation, this causing their slipping relative to the wire and necessitating the increase of power of the drive motors.

One of the most effective methods to eliminate this phenomenon can be the use of mechanisms with the large diameter feed roller and several pressure rollers of a smaller diameter [1]. In such a mechanism the roller diameter does not influence the electrode wire feed rate, and does not create any additional load in terms of the moment of resistance to its feed. We would think that such systems should be introduced into the practical design work on a broader scale.

Problems of development of clamping devices for electrode wire feed mechanisms are also urgent. A device with calibration of the clamping force, for instance, series of PSh107V semi-automatic machine is preferable [2].

Feed mechanisms with planetary rollers have certain technical and technological advantages (lowering of resistance at electrode wire feed in the guide; stabilization of the welding process; reduction of the semi-automatic machine overall dimensions; increased accuracy of the wire guiding to the butt due to its straightening), and they have become accepted in semi-automatic machines and robotic welding systems [3]. However, application of planetary mechanisms is associated with some difficulties, namely electrode wires of a practically ideal geometry are required; a finely-dispersed powder of the wire material is formed, which clogs the guide. These problems are partially solved by using a mechanism with planetary rollers, mounted on the crossing and additionally inclined axes (Figure 2). There also exist other technical and technological solutions, providing the reliability of operation of the above mechanisms.

Of special interest are mechanisms, performing electrode wire feed with a controlled non-stationary nature of the motion. Mechanisms with an intermittent and modulated electrode wire feed operate in the frequency range from several fractions to units of Hertz. Pulsed feed of electrode wire is characterized by the frequency range of several tens of Hertz and quite large values of acceleration, which should be controlled [4].

As regards electrode wire feed in welding with mode modulation, more efficient is the use of engineering solutions for controlling the frequency of rotation of the motor drive shaft, allowing for the inertia and frequency characteristics of the electric drive.

Intermittent nature of the electrode wire motion can be implemented by different methods, but all the existing diversity of engineering solutions is reduced

**Figure 2.** Planetary mechanism of electrode wire feed

to roll knobbing of the feed rollers [5] (being the optimum variant based on the experience of the E.O. Paton Institute Experimental Design-Technological Bureau), use of additional devices, interrupting the electrode wire feed (roller withdrawal and pressing down, creating a wire stock, or reciprocal motion of the feed mechanism to the welding zone, etc.).

We attribute the great diversity of engineering solutions for pulsed electrode wire feed to the developers desire to simplify them, while providing certain technological advantages. The simplest of the existing ones are the mechanisms of pulsed feed with a one-sided grip. They widely use devices, in which the grips make a reciprocal motion. Less known are the mechanisms, in which the grips are stationary, and the wire moves between them in a pulsed manner at forced oscillations.

Figure 3 gives an example of a successful implementation of such an engineering solution. Proceeding from the gained experience, we believe that the mechanism with stationary grip elements is more efficient in terms of jamming safety, and independence of the wire feed step on the force of its resistance to the motion, compared to the moving one-sided grips.

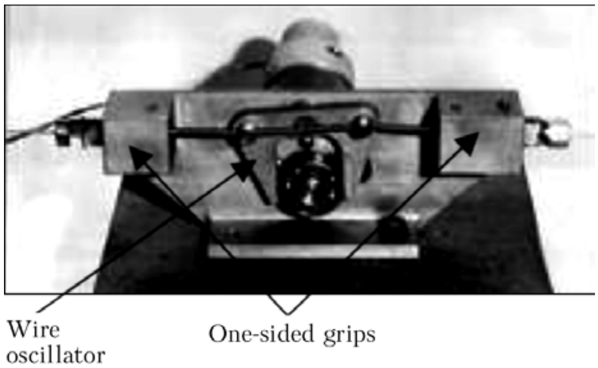


**Figure 3.** Feed mechanism with stationary grips

**Figure 4.** Electrode metal transfer onto a vertical plane at a combined action of the pulsed feed mechanism and pulsed power source [4] (obtained at filming by a camera)

Mechanisms of pulsed feed of electrode wire using one-sided grips, can be used in certain cases (solving local tasks, small-scope jobs, use of drives to perform intermittent motion, for instance, of electric magnets). Here the main efforts of the designers are aimed at developing reliable grips, accurately reproducing the specified motion parameters. A number of such developments have been made recently, and in the near future we can anticipate the introduction of new designs of electrode wire feed mechanisms, based on one-sided grips, for instance, by the type of those given in [6].

To solve the problems associated with provision of reliability of batch-produced equipment, effect on the technological process of welding, surfacing, repeatability of its results, it is more rational to use mechanisms with pulsed rotation of the feed roller. This new type of mechanisms with a quasiwave transducer (QWT) [7] enables regulation of accelerated motion of electrode wire for a controlled transfer of electrode metal. However, achievement of the desirable results is associated with difficulties of achieving the required parameters of acceleration. It is required to optimize the geometrical dimensions (limit the length) of the guide channel to avoid a change of the parameters of electrode wire motion, prevent slipping of the rollers, and increase vibration of the entire feeding assembly. In view of all that, we developed and suggested a process of welding with pulsed feed of electrode wire, at which the mechanical pulsed action is complemented by the action of an electric pulse from the arc power source by certain algorithms. Both the pulsed feed mechanism, and impulse power source operate in a sparing mode in terms of power consumption. The required control of electrode metal transfer is achieved with the selected algorithms of such a joint action.

A typical pattern of electrode metal transfer in mechanized welding in $CO_2$ with Sv-08G2S type wire of 1.2 mm diameter on the vertical plane with certain algorithms of the pulsed feed mechanism with QWT and power source with a sinusoidal pulse generator, is shown in Figure 4. Investigation of welding with combined control leads to the conclusion of its considerable influence on the arc welding process.

The above examples clearly show that in order to achieve complete control of the process of mechanized arc welding, it is necessary to more thoroughly select the system and mechanism of electrode wire feed, and to achieve a greater effect, combined solutions are possible using controlled non-stationary processes that enhance the effect.


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5. Duben, Z. *Device for intermittent wire feed in arc welding.* Pat. 116243 ChSSR. Int. Cl. M 05 B. Publ. 15.10.65.
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# THESIS FOR SCIENTIFIC DEGREE

**E.O. Paton Electric Welding Institute of the NAS of Ukraine**

**Yu.E. Rudoj** (PWI) defended on 1st of March 2006 candidate's thesis on topic «Development of Gradient Heat-Protection Coatings and Electron Beam Technology for their Deposition on Gas Turbine Blades».

In the work conditions of deposition on a substrate of heat-protection coatings with gradient of chemical composition and structure over their thickness with external ceramic layer are determined, which include electron beam heating and evaporation of the mixture of metals (alloys) and oxides with different vapor pressure at the evaporation temperature in the form of a compressed pallet located on the end of a ceramic ingot from stabilized zirconium dioxide.

Results of investigation of the structure and functional properties of gradient heat-protection coatings of metal–ceramics with transition zone on the basis of the systems Al–$ZrO_2(Y_2O_3)$, Al–Pt–$ZrO_2(Y_2O_3)$, and Al–Y–$ZrO_2(Y_2O_3)$ are presented, which were produced by single-stage application process. Possibility of regulating structure of gradient coatings by means of chemical composition change of the mixtures being evaporated is shown. Optimization of chemical composition of the pallet being evaporated allows obtaining long-term gradient heat-protection coating with high thermal-cycle durability in air.

The dissertator studied regularities of chemical composition and structure change of the transition zones metal–ceramics of the deposited gradient coatings as a function of the deposition process technological parameters.

Optimum conditions of deposition within one technological cycle of the metallic binding layer, gradient transition zone and external ceramic layer, and conditions of subsequent heat treatment in vacuum of gradient heat-protection layers are developed.

Mechanism of formation of gradient structures obtained by electron beam evaporation of zirconium dioxide-base composite ceramic ingot is considered. Results of investigation of chemical composition, structure, and properties of recommended heat-protection gradient coatings NiCoCrAlY + AlCr / $ZrO_2(7Y_2O_3)$ and $Me_xC_y$ + NiAl / $ZrO_2(7Y_2O_3)$ deposited from vapor phase on surface of high-temperature alloys are presented.

Thermal-cycle tests of standard and gradient heat-protection coatings applied on various high-temperature alloys were carried out, which showed advantage of gradient heat-protection coatings with a binding layer of $Me_xC_y$ + NiAl and NiCoCrAlY + AlCr types due to their higher thermal stability as a result of formation of gradient transition zones on the boundaries binding layer / high-temperature alloy and binding layer / ceramic layer.

Single-stage electron beam technology based on evaporation of a composite ceramic ingot allows depositing heat-protection coatings on gas turbine blades with higher level of reliability and durability and lower cost of coatings in comparison with existing multistage technological processes of heat-protection coating application.

# PATENTS IN THE FIELD OF WELDING PRODUCTION*

**Method of renovation of worn surfaces of steel parts** characterized by the fact that loading of the part is performed with preliminary determination of the metal yield point and corresponding to it load, at which electric arc surfacing of the worn surface is performed and the load is removed, due to which tensile stresses in mentioned areas are reduced, and after natural cooling additional bending load is applied to the part with soaking and loading necessary and sufficient for formation of compression stresses in critical areas of the part being surfaced in which the highest probability of failure exists. Patent 2264280 RF. G.R. Arslanova, R.M. Salakhutdinov and K.S. Selivanov (Ufa SATU) [32].

**Device for welding with modulated current** characterized by the fact that the modulator is installed into break in the welding current control circuit between the automatic control system and the direct current source, whereby the modulator consists of a switch, the scheme of its control and a divider, the switch being made with one output, which is at the same time output of the modulator, and three inputs, one of which is a controlling one and connected to the switch control circuit output; the second input is connected to the automatic control system, and the third one to the divider. Patent 2264896 RF. S.R. Amanov, A.R. Shishkin, A.V. Kargin and D.Yu. Kopylov (AVTOVAZ Company) [33].

**Method of arc welding with activating material**, in which into the arc burning zone a strip manufactured from neutral in relation to the metal being welded material is fed, characterized by the fact that on the strip surface an activating material is applied, which consists of the mixture of the activating flux and a polymer with the following ratio of components, wt.%: 5--80 of activating flux; 20–95 of polymer. Patent 2264897 RF. S.G. Parshin and Yu.V. Kazakov [33].

**Method of pressure welding of dissimilar metals in air** characterized by the fact that compression force is not maintained in the process of welding, the butt area is heated by a loop inductor up to the welding temperature, after which heating is switched off for 1--2 s and an additional heating pulse is supplied up to the welding temperature ---- 0.8–0.9 of melting point. Patent 2264898 RF. A.V. Gubarev, V.V. Gubarev, V.V. Gubarev and V.I. Yudakov [33].

**Device for renovation of parts by electroslag cladding** characterized by the fact that a weighing mechanism is made in the form of an equilibrating cable-block system that includes a pillar, on which a guide is located connected by means of a cable with a device for attachment of a chill mould and a part and with an equilibrating load. Patent 2264899 RF. V.V. Vashkovets (Khabarovsk STU) [33].

**Device for renovation of parts by electroslag cladding** containing a drive for vertical movement of electrodes, which consists of a screw and a load nut connected with a vertically moving carriage located on the guide column, characterized by the fact that the load nut is detachable and additionally contains a mechanism for withdrawing it from contact with the screw, and the carriage is connected through the cable-block system with a counterweight. Patent 2264900 RF. V.V. Vashkovets (ditto) [33].

**Method of light beam welding of materials of different thickness and dissimilar materials** characterized by the fact that parts having different thermal properties or thicknesses are welded by a light beam of lazar radiation, on which two glasses with different thermal properties are installed. Patent 2265901 RF. A.P. Budnik, E.V. Mishchenko and D.P. Sokolov (Voronezh STU) [33].

**Power source for arc welding** characterized by the fact that the first and the second secondary windings consist of three sections, of which the third one is common for the windings, whereby the first secondary winding consists of the first and the third sections connected in series and in accordance, and the second secondary winding consists of the second and the third sections connected in series and in accordance, one end of each section being connected to the common point and unconnected ends of three sections being connected to diagonals of alternative voltage of two bridges consisting of six gates, two of which are common for two bridges. Patent 2265504 RF. O.I. Sakhno, L.I. Sakhno, P.D. Fyodorov and A.I. Komarchev (St.-Petersburg SPU) [34].

**Method of welding of item from nickel-base superalloy**, in which the whole zone of a weld and the area adjacent to the weld zone of the item is heated up to the temperature of maximum ductility, which is higher than ageing temperature and lower than initial melting point of the mentioned superalloy, and such temperature is maintained during welding and solidification of the weld, temperature of the welded item is increased up to the temperature of removal of the mechanical stresses and the welded item is cooled down to the temperature below disperse solidification range of primary gamma-phase at the rate, which is efficient for reducing precipitation of primary gamma-phase. Patent 2265505 RF. M. Foster and K. Updegrove (Chromealloy GAS Turbine Corporation, USA) [34].

**Method of manufacturing electrodes for resistance spot welding**, in which a die is filled with the electrode material and stamped with a punch, characterized by the fact that the electrode material, consisting of spent electrodes and 0.2–0.5 wt.% Cr, is molten up to the melting point and after cooling down to the temperature 900--950 °C it is stamped by a punch and then quenching of the electrode takes place. Patent 2265506 RF. A.P. Rukosuev, Yu.G. Novoseltsev, O.A. Rukosuev and D.V. Antonov (Krasnoyarsk STU) [34].

*The information about patents published in the bulletin of Russian Federation «Inventions. Useful models» for 2005 is presented (in square brackets the bulletin No. is indicated).

**Method of thermo-mechanical cutting**, in which a directed supersonic jet of solid fuel combustion products acts on the material, characterized by the fact that between charges of solid fuel a pyrolizing material is placed and a layer of the same material is applied on surface of the nozzle block wall, which directs the jet, whereby the pyrolizing organic material, the content of which is not more than 10 % of solid fuel mass, is used that includes 2–5 wt.% of metal powders with size of particles 50–250 μm and density 5–8 g / $cm^3$. Patent 2266178 RF. G.V. Kuznetsov and T.N. Nemova (SRI of Applied Mathematics and Mechanics of Tomsk STU) [35].

**Robot for electric arc welding** of the rail zone containing a torch for electric arc welding that can move for performing welding of a rail and a movement control means, characterized by the fact that the movement control means includes the first means for controlling movement of the electric arc welding torch over perimeter of the subject to welding zone in such way that geometric places of welding in the mentioned zone be able to be read by the reading means connected with the memory means for entering into the memory data on mentioned geometric places of welding obtained on the rail in the course of reading, and the second means for automatic movement control of the torch for electric arc welding along subject to the welding zone within the limits of mentioned reading and as a function of geometric places of welding according to stored in the memory data. Patent 2266179 RF. J. Soron and J. Gohn (Societe Jozef Soran Materiele Industriel Z.I. Le Bord, France) [35].

**Device for feeding of rotating surfacing wire** characterized by the fact that in the housing of a wire feeding mechanism over the axis of the wire feeding channel aligning bushings are installed in the amount $2n + 1$, where $n = 1, 2, 4, 8$, whereby the bushings are located at the same distance from each other. Patent 2266180 RF. N. Mashrabov (Chelyabinsk STU) [35].

**Mechanism for pulsed feeding of welding wire** characterized by the fact that the converter of constant feeding into the pulse one is installed at the outlet from the welding hose and consists of an electric magnet with a pusher mechanism on the armature and a ball grip which has a spring-loaded stop and is made separately from the electric magnet, whereby clearance between the pusher mechanism and the stop of the converter can be regulated. Patent 2266181 RF. O.G. Brunov, V.T. Fedko, A.V. Kryukov et al. (Tomsk Polytechnic University) [35].

**Device for seam ultrasonic lap welding of pipe edges** containing a concentrator, a roller fixed at the concentrator end, the other end of which is connected with a vibrator, characterized by the fact that the concentrator is made in the form of a stage waveguide, whereby diameter of the first stage of the waveguide for the end of the concentrator designed for connection with the vibrator is bigger than diameter of the second stage of the waveguide for the end of the concentrator, on which the roller is fixed, diameter of the roller being selected equal to the diameter of the waveguide second stage at the concentrator end. Patent 2266182 RF. E.Yu. Bukharev, A.M. Mirzoyan and N.N. Ryzhov (NPP «Mayak-93» Ltd.) [35].

**Method of producing flat multilayer billets and sheets from dissimilar titanium alloys** characterized by the fact that rolled sheet interlayers from commercially pure titanium are used, whereby diffusion bonding of the package layers and interlayers is performed in the vacuum chamber of hydraulic press at the vacuum pressure of the chamber space $(6.67–13.3)\cdot 10^3$ Pa, temperature of the package heating equal to (0.53–0.6) of titanium melting point, ratio of mean pressure of the package compression to titanium yield point (1.5–3.0), and duration of exposure of the package to the pressure for 1–2 h. Patent 2266183 RF. N.D. Lukashin, Ya.L. Solomonik and A.Ya. Vinokurov (Moscow SHMI) [35].

**Method of assembly of a jig for assembly-welding of automobile bodies** including installation on the guide rails by basic elements of a central part of the jig with installed and adjusted jig units consisting of jig plates and working blocks, attachment of the jig central part by clamps, bringing of side door bodies nearer, and adjustment of mutual fixing units. Distinctive features are described. Patent 2266184 RF. V.A. Nosov and A.E. Khabarov (GAZ Company) [35].

**Method of single-piece connection of parts** characterized by the fact that a brazing filler alloy in the form of nano particles is used, whereby the surfaces being connected are preliminary processed up to the maximum roughness degree $10^3$ nm. Patent 2266801 RF. F.P. Demidov, V.B. Lap-shin, A.A. Palej and M.Yu. Yablokov [36].

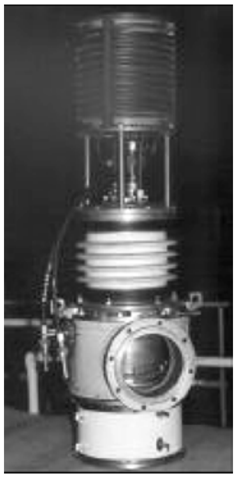
**Device for resistance laser treatment** containing an optic laser head, a probe, and a sensor of movement, characterized by the fact that the sensor is made in the form of a source of digital signals comprising two spring-loaded relative each other discrete plates, and the probe is equipped with adjustable over height rod, by means of which it is connected with one of the plates, whereby one of the plates is spring-loaded by installed on the rod springs of back action, and the other plate is spring-loaded by the installed on the head lead screw. Patent 2266802 RF. S.V. Usov, Yu.A. Belobratov and M.M. Kokoulin (Tulamashzavod Company) [36].

## ELECTRON BEAM MELTING OF TITANIUM ALLOYS

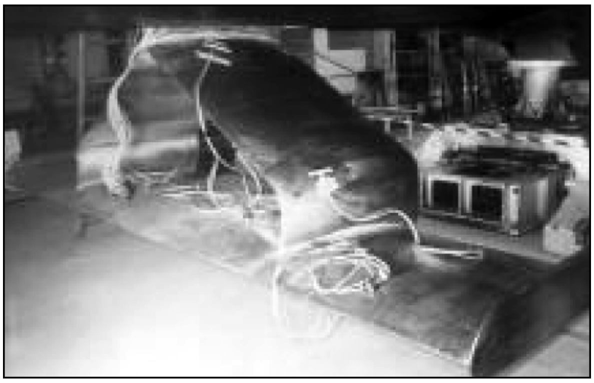
Fundamental research performed at the E.O. Paton Electric Welding Institute of the processes of alloy component evaporation from the melt in vacuum and ingot solidification at electron beam melting with an intermediate crucible (EBMIC) allows forecasting the composition and structure of the produced titanium alloy ingots and producing ingots of a guaranteed composition. Application of an intermediate crucible eliminates penetration of high- and low-density inclusions into the mould.

EBMIC mini-plant of 3000 t / y capacity

Appearance of UE-121 electron beam unit



«Paton-300» axial gun



Underwater wing of a sea boat of titanium alloy PT-3V

Aircraft parts (VT22 alloy) made of EBMIC ingots

EBMIC ingots of VT6 titanium alloy of 400 mm diameter

The sequence of and special fixture for continuous feeding of alloying elements into the charge have been optimized. Technology of producing ingots of alloyed titanium alloys (VT6, PT-3V, VT22, etc.) in electron beam unit UE-121 has been introduced fitted with axial guns «Paton-300» of the rated power of 300 kW with differential pumping, thus allowing the melting process to be conducted in a stable uninterrupted mode. The produced ingots fully meet the requirements of international standards both as to the content of alloying elements and impurities, and as to their distribution over the ingot cross-section.

**Application.** The developed technology can be used for a guaranteed production of high-quality ingots of titanium alloys, not containing any low- or high-density inclusions.

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