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EFFICIENT WELDING OF 80 mm S355ML PLATES USING A HYBRID LASER-ARC AND NARROW-GAP SAW PROCESS

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ABSTRACT

A new welding approach combining hybrid laser-arc welding (HLAW) and narrow-gap submerged arc welding (SAW) was investigated for joining 80 mm thick S355ML steel plates used in offshore wind turbine structures. The U-shaped joint preparation consisted of a 40 mm root face welded by HLAW, followed by a 23 mm narrow-gap section completed with multi-layer SAW passes. Process efficiency and mechanical performance were evaluated in comparison with conventional multi-pass SAW. The results showed that the combined process significantly reduces weld volume, filler metal consumption and heat input while maintaining the strength and toughness required for offshore applications. Mechanical testing confirmed a sound joint with a favorable and uniform hardness profile and adequate low-temperature performance. Charpy V-notch tests at -40 °C yielded average absorbed energies of $138\pm 45\text{ J}$ in the arc-dominated region and $65\pm 12\text{ J}$ in the laser-dominated region of the hybrid weld. The proposed approach provides an efficient and technically feasible solution for the fabrication of thick-walled offshore structures.

KEYWORDS: hybrid laser-arc welding, submerged arc welding, S355ML, thick-section steel, microstructure, mechanical properties, impact toughness

INTRODUCTION

The global transition towards sustainable energy generation has accelerated the deployment of wind energy systems. In line with the objectives of the Paris Agreement and international net-zero strategies, offshore wind power has emerged as a key technology for reducing greenhouse gas emissions while ensuring long-term energy security [1]. The continuous increase in turbine size, installation depths, and overall capacity demands has intensified the need for efficient fabrication technologies for large offshore structures.

Offshore wind turbines are predominantly constructed from thick-walled steel components such as towers, monopiles, and transition pieces. These cylindrical sections are manufactured from heavy plates that are rolled, tack-welded, and joined by longitudinal and circumferential welds. Thermomechanically rolled or normalized fine-grain structural steels including S355NL, S355ML, S420ML, and S460ML are commonly used according to EN 10025-3 and EN 10025-4. The indexes “N” and “M” denote normalized/normalized-rolled and thermomechanically rolled delivery conditions, respectively, while “L” indicates guaranteed low-temperature toughness. Typical impact energy verification temperatures are -20 °C

for N-grades, -50 °C for NL-grades, and -40 °C for ML-grades, reflecting the harsh environmental conditions offshore. These steels exhibit yield strengths between 355 MPa and 460 MPa and are commonly used for plate thicknesses ranging from 30 mm to 100 mm in tower and transition-piece fabrication, and up to 150 mm in monopile structures [2–4].

Welding such heavy sections presents significant challenges. Conventional multi-pass gas metal arc welding (GMAW) or SAW procedures require high heat input, large weld volumes, and long production times. These factors increase distortion and may negatively influence heat-affected zone (HAZ) properties. In thermomechanically rolled steels, excessive heat input can lead to grain coarsening and reduced impact toughness [5–7]. Maintaining consistent weld quality is essential for offshore fabrication, and welding procedure specifications (WPS) must be followed rigorously. According to DNVGL-OS-C401 and DIN EN 10225, the allowable heat input for fine-grain structural steels typically ranges between 3.5 ± 0.2 and $5\pm 0.2\text{ kJ/mm}$, depending on grade and toughness requirements.

Although SAW remains the industrial standard for joining thick plates, either in single-, tandem-, or multi-wire configurations, it still involves large groove volumes and high filler metal

consumption. Modern high-productivity concepts such as narrow-gap SAW (NG-SAW) and narrow-gap GMAW (NG-GMAW) processes aim to address these limitations. Narrow-gap techniques significantly reduce weld volume, while still enabling reliable fusion even in sections exceeding 200 mm [8–10].

Laser beam welding (LBW) is another high productivity joining method offering excellent penetration capability and very high travel speeds. However, the extremely rapid cooling associated with LBW can lead to increased hardness and reduced toughness in both the weld metal (WM) and the HAZ, primarily due to martensitic transformation at high cooling rates [11, 12]. The analyzed literature shows that in LBW of Q345 high-strength low-alloy (HSLA) steel, moderate preheating of around 100 °C only marginally reduces cooling rates and partially suppresses martensite formation. A significant effect is achieved only at elevated preheating temperatures around 200 °C, where martensitic layers disappear completely, hardness becomes more uniform, and both fatigue performance and corrosion resistance improve noticeably [13]. However, such high levels of preheating are generally not economical for large-scale offshore structures due to the high energy demand and extended preparation times involved. Furthermore, LBW has limited tolerance for variations in joint gaps. Even with beam oscillation, gaps of only about 0.5 mm can be reliably bridged [14]. This narrow tolerance range restricts the applicability of LBW to components with larger dimensional deviations or heavy-wall sections.

Hybrid laser-arc welding (HLAW) combines the advantages of laser welding and arc welding within a common molten pool, thereby improving gap tolerance and weld pool stability while enabling deep penetration [15]. Studies on plate thicknesses up to 20–30 mm have demonstrated high process efficiency and sound mechanical performance [16–18].

Attempts to extend laser-based welding methods to thicker plate sections, such as welding under reduced pressure [19, 20], the use of electromagnetic weld pool support [21, 22], or the application of multi-layer LBW techniques [23, 24], have demonstrated that these approaches are technically feasible. However, despite ongoing technological progress and the availability of pilot systems [25, 26], their industrial applicability remains limited because of the high complexity of the equipment and the strong sensitivity to joint preparation.

For plate thicknesses up to 35 mm, combined LBW and SAW processes have been successfully implemented [27]. This method combines two welding techniques within a shared molten pool, enabling thick plates to be joined using double-sided welding.

A practical and industrially feasible alternative is the sequential application of HLAW for the root region followed by SAW filling passes. The HLAW parameters are intentionally selected to avoid full penetration, eliminating the need for root backing or root face formation and thereby increasing the robustness and ease of implementation of the process. The subsequent SAW layers are applied from the opposite side and overlap the hybrid weld root, thereby completing the joint cross-section. Previous work on 30 mm EH36 plates [28] demonstrated that such process combinations ensure complete fusion, stable overlap between weld regions, and adequate mechanical performance even when the joint edges are non-ideal or produced by plasma cutting. Scaling this approach to significantly thicker materials requires an adapted groove geometry and a carefully selected set of welding parameters to ensure reliable metallurgical intersection between the hybrid root pass and the subsequent SAW weld layers.

The present work investigates a combined HLAW and SAW welding approach for 80 mm thick S355ML plates. The study focuses on evaluating process efficiency, weld integrity, microstructural evolution, hardness distribution, and low-temperature toughness to assess the suitability of this hybrid technique for offshore wind turbine fabrication. The specific objectives are to:

- establish a stable HLAW parameter window for deep but partial penetration;
- ensure complete metallurgical overlap between HLAW and NG-SAW;
- characterize the hardness profile across the weld and HAZ;
- evaluate the low-temperature toughness relative to offshore requirements.

MATERIALS AND METHODS

Plates made of S355ML were used for the welding trials. This thermomechanically rolled, fine-grain structural steel is defined in EN 10025-4 and offers a combination of high strength, good weldability and reliable low-temperature toughness. This makes it a suitable choice for welded, load-bearing structures in offshore environments. Its low carbon equivalent (CEV), calculated in accordance with EN 1011-2, contributes to the favorable weldability of the material and minimizes the risk of hydrogen-assisted cold cracking. The chemical composition and mechanical properties of the steel are listed in Tables 1 and 2, respectively.

In the GMAW component of the HLAW process, a 1.2 mm solid wire of type Union K 52 Ni (EN ISO 14341-A: G 50 6 M21 Z3Ni1) was used, together with an argon–carbon dioxide shielding gas contain-

Table 1. Chemical composition of S355ML steel plates according to the manufacturer’s material certificate.

Element, wt.%										CEV, %
C	Si	Mn	P	S	Ni	Cr	Mo	V	Fe	
0.046	0.056	1.42	0.009	0.0015	0.067	0.036	0.008	0.001	Bal.	0.31

ing 18 % CO₂ at a flow rate of 25 L/min. For the SAW process, a 4 mm solid wire electrode of grade BA S2 in accordance with EN ISO 14171-A was selected, used in combination with the agglomerated aluminate-basic flux BF3.5. To ensure adequate toughness, the SAW consumable manufacturer recommends limiting the heat input to approximately 2 kJ/mm.

For the HLAW experiments, a 60 kW laser system from TRUMPF was employed. The system was configured by combining three 20 kW TruDisk lasers to achieve the required total power. The laser power was transmitted through the optical combination of three 100 μm fibers into a single process fiber with a core diameter of 300 μm. The TruDisk lasers operated at a wavelength of 1030 nm. A modular laser processing head of the type MPH-EHP (PT Photonic Tools GmbH) was used as the laser optics. The optical setup had a focal length of 400 mm and a magnification ratio of 2:1, resulting in a laser spot diameter of approximately 600 μm in the focal plane.

The GMAW was operated using a QINEO power source (CLOOS GmbH) in DC+ spray arc mode. The hybrid welding head, comprising the laser optic and a fixed GMA torch, was mounted on an industrial robot that guided the process along the joint.

The experiments were performed under constant geometric conditions, with the GMA torch inclined at $\gamma = 25^\circ$ to the laser axis, a laser focus position of $\Delta z = -7$ mm, a wire stick-out of 18 mm. The parameters such as laser power (P_l), arc power (P_{arc}), and welding speed (v_w) were varied to obtain optimal welds with maximum penetration depth and high external as well as internal weld quality. The HLAW process configuration is presented in Figure 1.

Table 2. Mechanical properties of S355ML steel plates according to the manufacturer’s material certificate.

R_e , MPa	R_m , MPa	A_5 , %	KV , J	Hardness, HV10
401	484	29	360 (−40 °C)	160

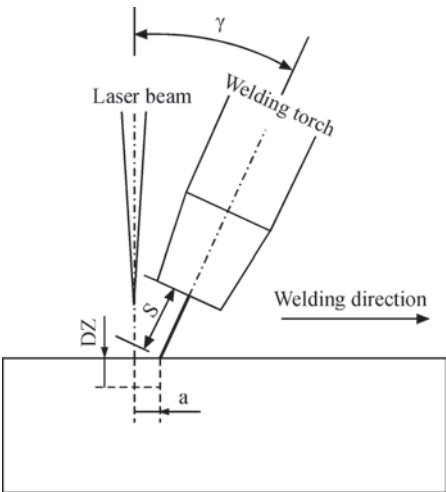


Figure 1. HLAW process configuration

The SAW experiments were performed on an industrial longitudinal welding system capable of operating with up to five wires, each powered by a PERFECTarc® 1500 AC/DC current source from SMS group GmbH. Although the facility allows multi-wire operation with a combined current of up to 7500 A, only a single wire was used in this study to keep the heat input within the required limits.

For the welding trials, S355ML plates with dimensions of 500×200×80 mm³ were prepared. The specimen size was selected considering the available handling and positioning capabilities in the laboratory.

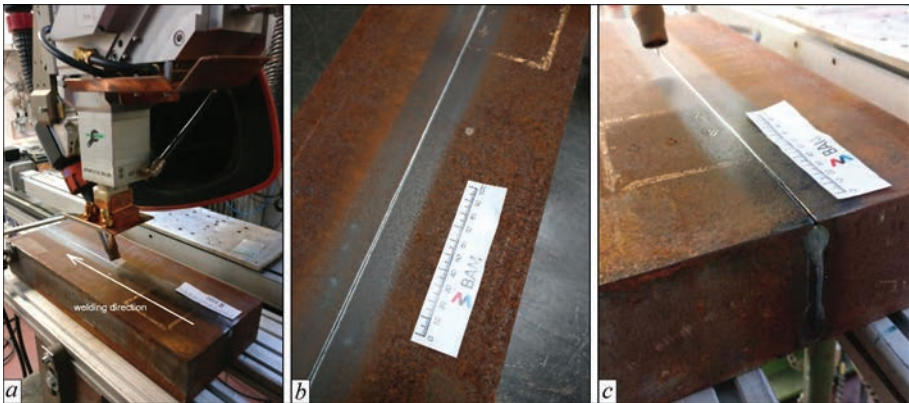


Figure 2. HLAW system: a — experimental setup; b — 80 mm thick specimen prepared for HLAW root pass

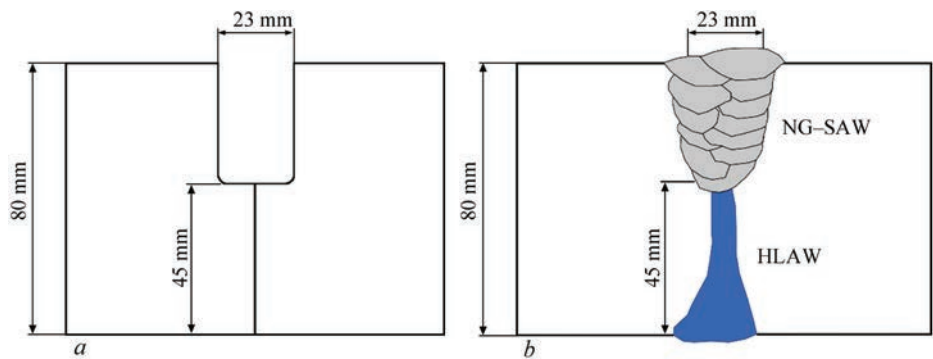


Figure 3. Schematic illustration of the weld joint in an 80 mm thick plate: *a* — joint preparation with U-groove and root face; *b* — weld produced by the combined HLAW and NG-SAW process

The experimental setup for the HLAW trials and the prepared welding specimen are shown in Figure 2, *a* and *b*, respectively.

After the HLAW root pass was applied to the butt joint, a U-shaped groove with a width of 23 mm was machined from the opposite side. The HLAW parameters were selected to achieve approximately half of the 80 mm plate thickness, producing a hybrid weld depth of about 40 mm and defining the remaining depth of the U-groove. This groove was subsequently filled by applying NG-SAW layers. A schematic illustration of the U-shaped groove and resulting weld is shown in Figure 3, *a* and *b*, respectively.

After welding, the specimens were subjected to mechanical and metallurgical testing. The testing program included Charpy impact tests in accordance with EN ISO 9016, light-microscopic examination of metallographic cross-sections in accordance with EN ISO 17639, and Vickers hardness measurements in accordance with EN ISO 9015-1.

RESULTS

WELDING EXPERIMENTS

To produce the 80 mm thick weld, the HLAW process was first applied to the butt joint. The welding parameters were selected to achieve a penetration depth of approximately 40 mm. These parameters were determined in preliminary HLAW trials, which were conducted to establish the laser power required to reach this depth. A target penetration depth of 40 mm was achieved at a laser power of 45 kW, while even at the maximum available power of 60 kW the penetration increased only slightly to approximately 42 mm. The

welding speed was kept constant at 0.6 m/min. This saturation trend indicates that further increasing the penetration depth under the given conditions is not technologically justified, as it would require a disproportionate rise in laser power.

Based on these findings, the following parameters were selected for welding the 80 mm plates: $P_l = 45$ kW, $v_w = 0.6$ m/min, and $v_f = 12$ m/min. At this wire feed speed, the GMAW operated at an arc voltage of $U_{arc} = 32$ V and a current of $I_{arc} = 300$ A. The HLAW process power was therefore approximately 54.6 kW, corresponding to a heat input (E_{HLAW}) of about 4.4 kJ/mm, assuming an overall process efficiency factor of 0.8 for both heat sources.

In the second step, welding was carried out from the opposite side using the NG-SAW process. To prevent the torch from jamming in the narrow groove, a width of 23 mm was selected. To achieve proper remelting of the hybrid root and a sound overlap between the HLAW and SAW weld regions, the first SAW pass was welded using direct current DC+, which provides deeper penetration. The subsequent filling passes were deposited with alternating current (AC) to increase deposition efficiency. Process control was further optimized through digital adjustment of the power source, with the I-balance set to 30 % positive and 70 % negative, enabling higher deposition rates without increasing heat input. The welding parameters and number of passes for the NG-SAW process are summarized in Table 3.

The interpass temperature during the SAW fill passes was monitored using a Type K thermocouple connected to a temperature data logger Lascar EasyLog EL-USB-

Table 3. Parameters of the NG-SAW process

Weld passes	Number of passes	Current type and waveform	I_{arc} , A	U_{arc} , V	v_w , m/min	E_{SAW} per pass, kJ/mm
Root pass	1	DC+	600	33	0.6	1.8
Fill passes	9	AC, I-balance 30/70	600	35	0.6	1.9
Cap passes	2					

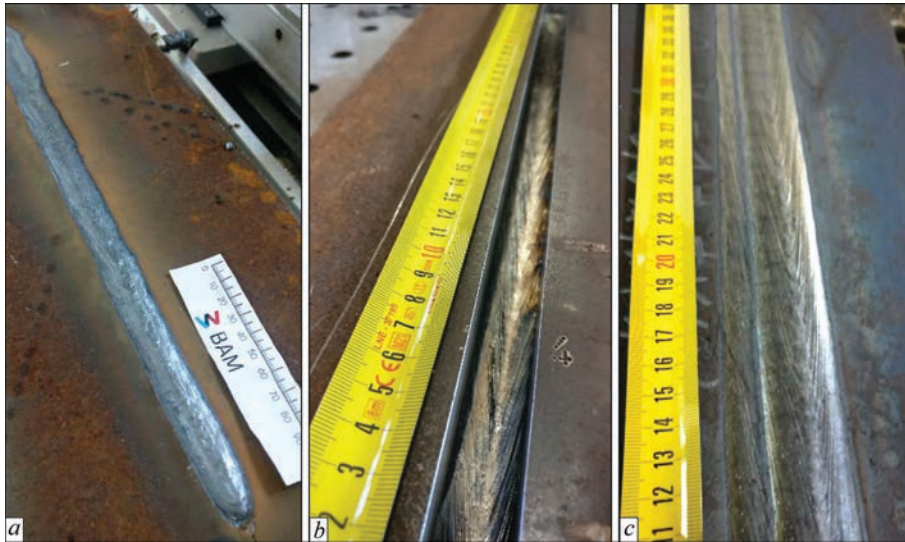


Figure 4. Welding of an 80 mm S355ML plate: *a* — HLAW root pass; *b* — NG-SAW filler layers, and *c* — fully welded joint

TC-LCD. The interpass temperature was maintained within a range of approximately 150 to 180 °C.

A representation of the individual welding steps, namely the execution of the HLAW root pass, the deposition of the SAW filling layers, and the fully welded joint, is shown in Figure 4, *a–c*, respectively.

Full cross-sectional macrographs in Figure 5, *a–c* confirmed that the combined HLAW–NG-SAW joint was free of cracks, lack-of-fusion defects, and macroscopic inclusions. Figure 5, *a* shows the HLAW weld produced in the first step, Figure 5, *b* illustrates the deposition of the NG-SAW fill passes, and Figure 5, *c* displays the completely welded 80 mm thick joint with the marked positions of the Charpy V-notches.

The transition between the hybrid and SAW regions was smooth, with no evidence of an unmelted interface or metallurgical discontinuities. The fusion boundaries along the U-groove walls were distinct and uniform, and the bead geometry exhibited a consistent convex profile without undercut.

HARDNESS MEASUREMENTS

The Vickers microhardness (HV1) results for the combined HLAW–NG-SAW weld are shown in Figure 6, *a* and *b*. Measurements were taken along three lines located in the SAW fill region, the mid-thickness of the HLAW weld, and the HLAW region approximately 5 mm below the plate surface (Figure 6, *a*). The corresponding hardness profiles for the three regions are shown in Figure 6, *b*.

The hardness profiles are uniform across all regions. The base material exhibits an average hardness of 164 ± 6 HV1, while the weld and HAZ regions show mean values of 172 ± 10 HV1 (Row 1), 176 ± 12 HV1 (Row 2) and 171 ± 15 HV1 (Row 3). No significant softening or hardness drops were observed, and all values fall within the typical range for thermome-

chanically rolled S355ML steel ($160\text{--}190$ HV1). A slight hardness peak of about 205 HV1 appears in the mid-thickness of the HLAW fusion zone, likely caused by locally increased cooling rates and the resulting finer solidification structure. This variation remains moderate and does not indicate brittle microstructural regions, confirming that the combined HLAW–NG-SAW process provides controlled thermal conditions and a stable hardness distribution across the weld.

CHARPY TOUGHNESS TESTS

Charpy impact tests were carried out at -40 °C to evaluate the toughness of the hybrid weld. Since the NG-SAW parameters correspond to standard conditions for S355ML steel, Charpy testing focused exclusively on the HLAW region, where varying thermal cycles may influence toughness. The Charpy impact tests were carried out on standard $10\text{ mm} \times 10\text{ mm}$ specimens prepared in accordance with ISO

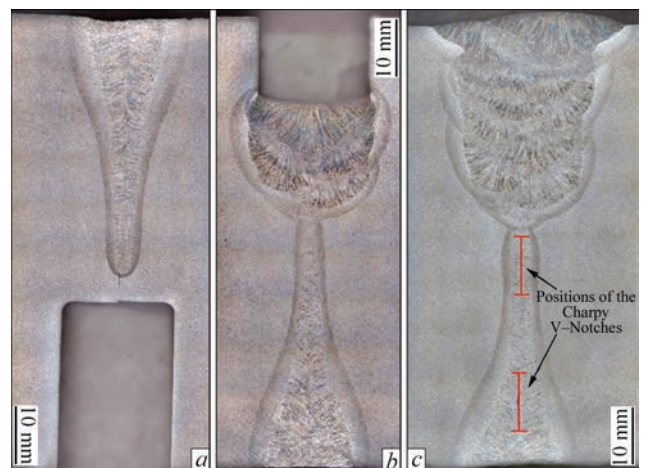


Figure 5. Macro sections of the 80 mm thick weld produced using the combined HLAW and NG-SAW process: *a* — HLAW weld from the first step; *b* — deposition of the NG-SAW fill passes in the U-groove; *c* — completed joint with indicated positions of the Charpy V-notch specimens

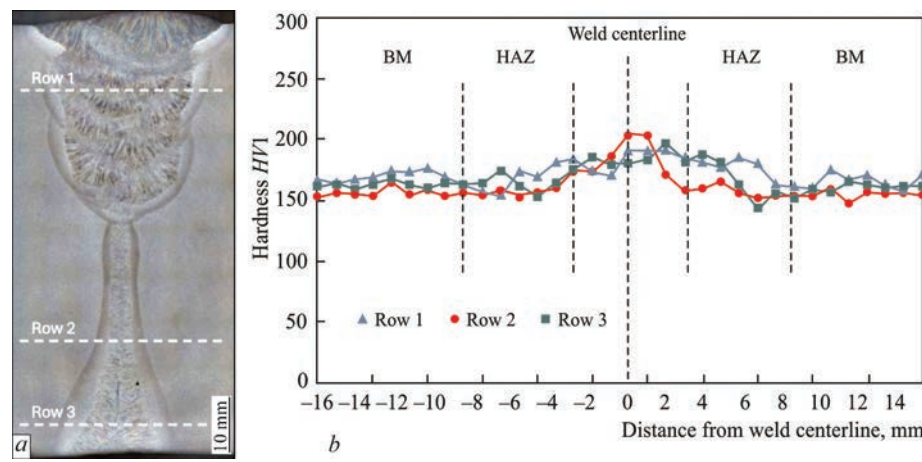


Figure 6. Results of Vickers microhardness ($HV1$) testing across the combined HLAW-NG-SAW welded joint in S355ML: *a* — schematic representation of the measurement lines; *b* — corresponding hardness profiles

148-1. Specimens were taken from two characteristic locations: the laser-dominated root region and the arc-dominated upper region of the HLAW weld, as it shown in Figure 5, *c*. Five specimens were tested in each case, with the V-notch positioned in the center of the WM. The results, presented in Figure 7, show a clear difference between the two weld regions.

The arc-dominated zone reached absorbed energies in the range between 90 and 188 J, with an average of 138 ± 43 J. The relatively large scatter reflects local microstructural variations of the arc-dominated part of hybrid weld, where the weld pool solidifies more slowly, leading to slight differences in grain size and toughness across the section. Nevertheless, all values are well above the minimum toughness levels generally required for structural steels in offshore applications at subzero temperatures.

The laser-dominated zone achieved lower values, between 52 and 80 J, with an average of 65 ± 12 J. This reduction is consistent with higher cooling rates and the formation of finer, harder microstructures. Despite the lower values, the toughness remains adequate for service conditions.

DISCUSSION

The combination of HLAW and NG-SAW proved to be a promising approach for joining 80 mm thick

S355ML plates. Compared with conventional multi-pass arc welding, the hybrid concept significantly reduces weld volume, filler metal consumption and overall production time, while maintaining the mechanical performance required for offshore applications. These results are consistent with previous investigations on HLAW of fine-grain steels of up to 20 mm thickness [17, 18, 30, 31], which reported high process efficiency and sound weld quality. The present study extends this concept to 80 mm plate thickness and demonstrates that a U-shaped groove with a root face of at least 40 mm enables the application of this combined process to substantially thicker sections.

In the proposed procedure, the HLAW pass is applied first, followed by NG-SAW from the opposite side using a multilayer sequence. The weld cross-section is completed by a stable overlap of the hybrid and SAW regions. A key advantage of this configuration is that no separate HLAW root formation is required, eliminating the need for mechanical or electromagnetic backing systems [26]. This underlines the robustness and practicality of the developed process. The combination of both welding techniques resulted in a substantial reduction in weld volume and heat input. As a result, the 80 mm joint could be completed in only 12 layers instead of the approximately 42 layers typically required for conventional 30° groove preparations [32].

Based on the measured weld cross sections, the material demand can be directly quantified per meter of weld length. For the combined HLAW-NG-SAW joint, the U-shaped groove with a cross-sectional area of approximately 805 mm² corresponds to a weld metal volume of 805 cm³ per meter. Assuming a steel density of 7.85 g/cm³, this results in a deposited weld metal mass of about 6.3 kg per meter of weld. In contrast, the conventional multi-layer SAW joint with a groove cross-section of 2355 mm² requires a weld metal vol-

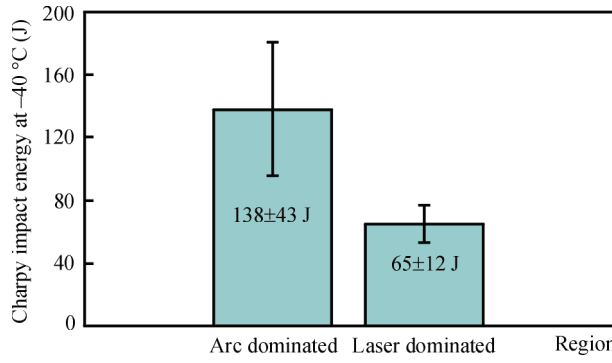


Figure 7. Results of the Charpy impact tests at -40 °C for two different regions of the HLAW weld

ume of 2355 cm³ per meter, which corresponds to approximately 18.5 kg of deposited weld metal per meter. Consequently, the combined HLAW–NG–SAW approach reduces the required weld metal mass by approximately 12.2 kg per meter, corresponding to a reduction factor of about 2.9.

Assuming identical welding travel speed of 0.6 m/min, the overall welding time per meter of joint scales linearly with the number of layers. For a weld length of 1 m, each pass requires approximately 1.67 min. Consequently, a conventional 42-layer SAW procedure results in an arc time of about 70 min/m, whereas the proposed 12-layer combined approach requires only about 20 min/m. This corresponds to a reduction in welding time by approximately 50 min/m, or a factor of 3.5, which represents a time saving of roughly 71 %, even before considering additional reductions in interpass and cleaning times.

The HLAW trials revealed a nonlinear penetration behavior. A penetration depth of approximately 40 mm was achieved at a laser power of 45 kW, while increasing the power to 60 kW yielded only a minor increase to about 42 mm. Similar saturation trends were reported by Kawahito et al. [33], who observed a maximum penetration of about 70 mm at 100 kW in austenitic stainless steel under comparable conditions. This suggests that the penetration limitation is caused by physical constraints, such as keyhole instability and energy losses induced by plasma. Therefore, process optimization should prioritize stabilizing the keyhole and minimizing the effects of plasma shielding rather than increasing laser power.

Heat input was a decisive factor in ensuring the required mechanical properties. According to EN 10225, an energy input of 3.5±0.2 kJ/mm is recommended for offshore steels, with a maximum of 5±0.2 kJ/mm permitted. The parameters applied in this study kept the total heat input below 5 kJ/mm, ensuring controlled thermal cycles throughout the joint.

The combined weld exhibited a uniform hardness distribution between 160 and 190 HV1. This is largely attributed to the tempering effect induced by the subsequent NG-SAW passes on the underlying HLAW weld. Comparable hardness profiles were reported for 30 mm EH36 hybrid welds, supporting the consistency and reproducibility of the process [28].

Charpy impact testing at –40 °C confirmed that the weld meets the toughness requirements for offshore steels. The arc-dominated region achieved an average absorbed energy of 138±45 J, while the laser-dominated region reached 65±12 J. The higher toughness in the arc-dominated region is attributed to the slower cooling rate and more homogeneous microstructure,

while the lower toughness in the laser-dominated zone corresponds to the increased hardness resulting from rapid solidification. Similar observations were reported by Üstündağ et al. [18] for 20 mm structural steels and by Volpp et al. [34], who linked short thermal cycles in laser-based processes to moderate reductions in toughness. Nonetheless, all measured values remain well within the limits specified by EN 10225 and DNV OS-C401, confirming the suitability of the hybrid joint for offshore applications.

Overall, the findings demonstrate that combining HLAW with NG-SAW provides a favorable balance between productivity and mechanical performance. The synergistic interaction of laser and arc enables deep penetration, efficient heat utilization and stable weld pool behavior, while the SAW passes ensure full fusion and a homogeneous microstructure. Whereas earlier studies focused on plate thicknesses up to 20–30 mm, the present work successfully applies the hybrid concept to 80 mm S355ML plates, highlighting its potential for industrial implementation in thick-walled offshore structures. The process satisfies key fabrication requirements by maintaining a controlled heat input, a uniform hardness profile and adequate impact toughness at subzero temperatures.

CONCLUSIONS

1. The combination of hybrid laser-arc welding (HLAW) and narrow-gap submerged arc welding (NG-SAW) has proven to be a reliable and efficient method for joining 80 mm thick S355ML plates. Compared with conventional multi-pass SAW using a 30° groove preparation, the combined approach reduces the weld volume by approximately 66 %, corresponding to a decrease in deposited weld metal from about 18.5 to 6.3 kg/m. In addition, the total welding time per meter of joint is reduced from approximately 70 min/m for a 42-layer SAW procedure to about 20 min/m for the 12-layer combined process, representing a time reduction of roughly 71 %.

2. The U-shaped joint design, consisting of an approximately 40 mm deep HLAW butt weld and a 40 mm NG-SAW section, enabled full joint completion without mechanical or electromagnetic backing, demonstrating the robustness and practicality of the proposed process concept.

3. Maintaining a nominal heat input below 5 kJ/mm ensured controlled thermal cycles and resulted in a uniform hardness distribution between 160 HV1 and 190 HV1 across the weld and heat-affected zone.

4. Charpy V-notch testing at –40 °C confirmed that the welded joint fulfills the toughness requirements for offshore applications, with absorbed energies of 138±45 J in the arc-dominated region and 65±12 J in

the laser-dominated region, both meeting the limits specified in EN 10225 and DNV OS-C401.

5. Overall, the developed HLAW-NG-SAW process satisfies the mechanical and technological requirements for thick-walled offshore structures, offering a favorable combination of high productivity, weld integrity and adequate low-temperature performance.

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DATA AVAILABILITY STATEMENT

The data presented in this study are available on request from the corresponding author.

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CONFLICT OF INTEREST

The Authors declare no conflict of interest

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RESTORATION OF RIVETED CYLINDRICAL TANKS USING WELDING

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ABSTRACT

As a result of the conducted research, it was found that low-carbon steel with content of 0.045–0.05 % C and 0.4–0.5 Mn was used for the shell of riveted tanks. Taking into account the nitrogen content at the level of 0.003 % N and oxygen 0.03 % O at a low carbon level, the studied steel may be considered as semi-killed. The sulphur content in steel can reach 0.1 % S. However, taking into account that the structure of the base metal and the HAZ is mainly represented by a sufficiently fine-grained ferrite, the impact energy of steel at $T = +15.5\text{ }^{\circ}\text{C}$ (60 F) satisfies the condition $KV_{\min} = 20\text{ J}$, embrittlement of the overheated zone metal is not observed. This indicates the possibility of sufficiently eliminating brittle fracture of welded joints of the specified steel and restoring the riveted tanks using welding. Based on the conducted studies, technical solutions for the complete (partial) replacement of riveted metal structures with welded ones were developed and tested during the repair of tanks after 114 years and 86 years of their operation: bottom, replacement of part of the shell, welding in pipes into the shell and installing a new roof. Successful operation for 10 years with the maximum level of oil product filling of the restored riveted tanks confirmed the correctness of the made decisions.

KEYWORDS: low carbon steel, weldability, brittle fracture, welded structures, riveted tank, repair

INTRODUCTION

One of the largest regions in terms of oil reserves and production is the South Caspian oil and gas basin, located on the Absheron Peninsula (Azerbaijan) and the adjacent waters of the Caspian Sea. Industrial oil production began there in the last third of the 19th century, and since 1870, more than 2 bln t of oil have been extracted [1]. The Nobel brothers made a significant contribution to the development of the oil industry in this region. After Robert Nobel discovered significant oil deposits on the Absheron Peninsula in 1873, he purchased the first oil refinery. In 1879, the Nobel Brothers Oil Production Company (abbreviated to “BraNobel”) was established. At that time, it was the first vertically integrated company in the Russian Empire [2], which developed rapidly. Thus, between 1879 and 1904, the company’s plants processed 19.6 mln t of oil and produced 6.2 mln t of kerosene, 39,000 t of petrol, 0.76 mln t of diesel fuel, 0.82 mln t of lubricating oil, and 11.7 mln t of oil residues. In 1889, Brown–Nobel reached its peak of 8.6 % from the global oil production [3]. Such rapid development of the company required the application of modern engineering ideas and technologies.

In 1877, a branch of the “Construction Office of Engineer A.V. Bari” was opened in Baku (Azerbaijan) [4]. V.G. Shukhov became the chief engineer. Together with A.V. Bari, they introduced their engineering developments to Ludwig Nobel, and after that, close cooperation between the companies began. The rapid development of oil refining and transportation re-

quired the creation of large-capacity tank farms. In 1878, V.G. Shukhov designed the world’s first steel riveted vertical cylindrical tank for storing kerosene, which was built under his supervision. In 1883, he published a theoretical justification for the design of a steel cylindrical tank [5]. By 1881, A.V. Bari’s office, along with the design and construction of oil pipelines, had built more than 130 steel tanks based on V.G. Shukhov’s designs. And between 1880 and 1905, under the leadership of V.G. Shukhov, about 3,000 riveted steel cylindrical tanks with a total capacity of 2.6 mln t were built for storing kerosene, petroleum products, oil residues, water, etc. [5].

As of 1990, many hundreds of riveted steel tanks were still in working order after 70–100 years of operation [2, 5]. Today, according to the authors’ estimates, up to 100 riveted tanks with a capacity of up to 10,000 m³ can be expected to be in operation in the region discussed above, so the issue of restoring their operability may be quite relevant.

DETERMINATION OF STEEL GRADE AND ITS WELDABILITY

The main factor determining the fundamental possibility of repairing riveted tanks is the grade of steel used for the cylindrical shell and the possibility of its repair welding to obtain a sufficient level of mechanical properties of welded joints to prevent brittle fracture of the tank. This issue requires particular study, given that riveted tanks were designed at a time when neither the theory of brittle fracture of steel structures [6] nor the determination of steel weldability existed.

The research was conducted on steel used for a riveted tank built in 1901 (project developer — Design bureau of the construction office of engineer A.V. Bari) and the tank built in 1929. Design developer is unknown. The characteristics of these tanks are shown in Table 1.

The studied joint was welded from 15 mm thick plates cut out from the lower course of the shell of the tank built in 1929 and simulated the welding of a vertical repair butt weld. The edge design in the welded joint was accepted symmetrical — X-shaped with an angle of 50°. For repair welding, a mechanized welding process 135 in a mixture M21–ArC–18 was selected using a solid wire G3Si1 with a diameter of 1.2 mm. The welding heat input per unit length for the root pass was $Q \approx 0.9$ kJ/mm, and for filling the root pass $Q = 1.1\text{--}1.2$ kJ/mm.

INVESTIGATION OF STEEL SHELL OF THE TANK BUILT IN 1901

For research purposes, sections of the riveted horizontal joint between the shell courses I and II were cut out (Figure 1). The rated metal thickness δ was 12 and 10 mm, while the actual thickness was 11.6, 11.6 and 10.3, 10.1 mm, respectively. Longitudinal specimens were used to study the mechanical properties of steel. The chemical composition of steel was determined by spectral analysis. The strength properties were determined based on the results of tests on short specimens with a working part diameter of 5 mm [7]. The impact toughness was determined on Mesnager

Table 1. Geometric characteristics of two steel riveted vertical cylindrical oil storage tanks

Year of tank construction	Diameter, m	Shell height, m	Shell courses thickness, mm
1901	23.4	9.54	12+11+9+8+7+6+5
1929	30.73	10.25	15+13+11+9+7+5+5

specimens (U-shaped notch) [8]. For steel thickness $\delta = 12$ mm, 10×10 mm specimens were tested, for $\delta = 10$ mm — 7.5×10 mm. Impact toughness was determined at temperatures of 0 and –20 °C [12].

Nonmetallic inclusions were studied during examination of unetched microsections, and microstructure was studied on microsections etched in sodium picrate. In order to compare the characteristics of steel under study with similar grades of carbon steel that had already been used for critical welded structures in the recent historical period, the evaluation of nonmetallic inclusions and steel structure characteristics was performed in accordance with GOST 1778–70 [9] and GOST 5640–68 [10].

The chemical composition of steel is shown in Table 2. Analysis of the obtained data shows that steel under study can be classified as low-carbon structural steel. Considering that steel was probably partially deoxidized with carbon and manganese, the silicon content may not be a decisive indicator of the degree of deoxidation. Taking into account the low nitrogen content $N = 0.0030$ %, which is lower than the permissible content $N < 0.008$ % for semi-killed (killed) steel

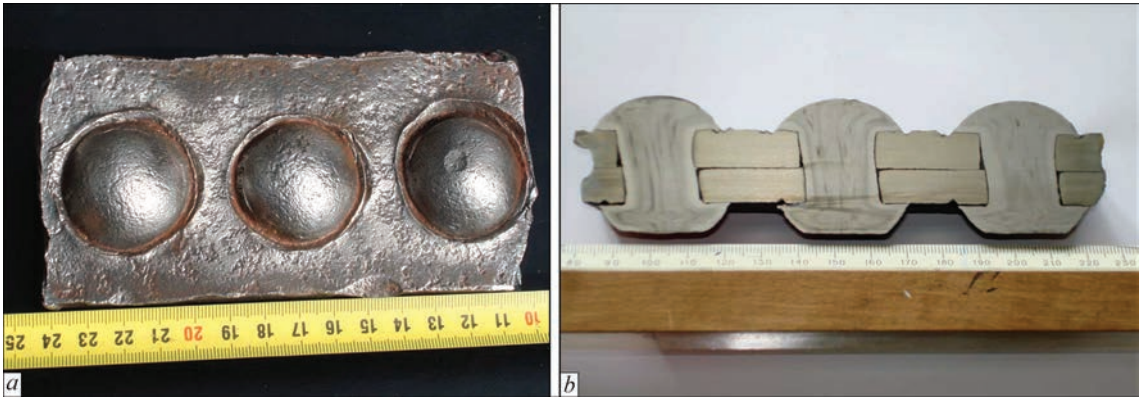


Figure 1. Horizontal riveted joint between the course I and II of the tank shell (steel thickness 12 + 10 mm), built in 1901: external view of the tank (a), transverse macrosection of the joint (b)

Table 2. Chemical composition of the steel in the shell of the tank built in 1901

Metal thickness δ , mm	Chemical element content, mass fraction, %						
	C	Si	Mn	S	P	O	N*
10	0.05	0.011	0.52	0.036	0.05	N/D	N/D
12	0.045	0.01	0.54	0.043	0.038	0.029	0.0030

*The nitrogen content in the rimmed steel is $N = 0.04\text{--}0.08$ %.

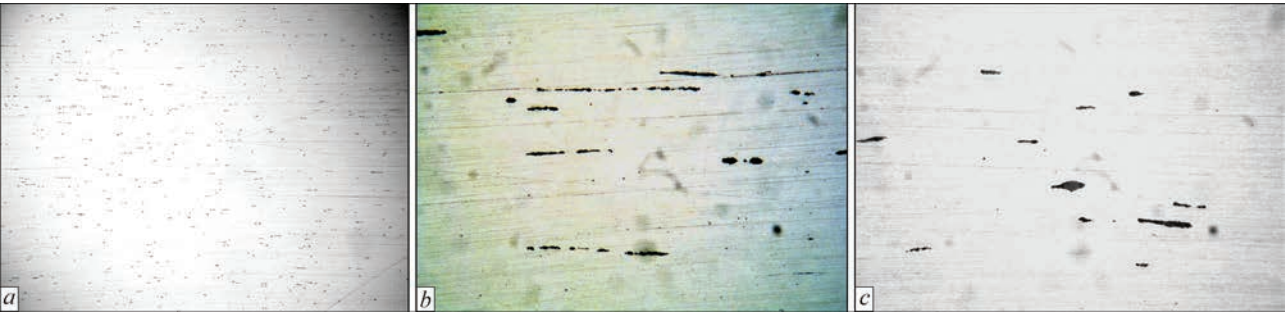


Figure 2. Manganese sulphides in the steel, $\delta = 12$ mm: *a* — $\times 100$; *b* — $\times 1000$; *c* — $\delta = 10$ mm, $\times 1000$

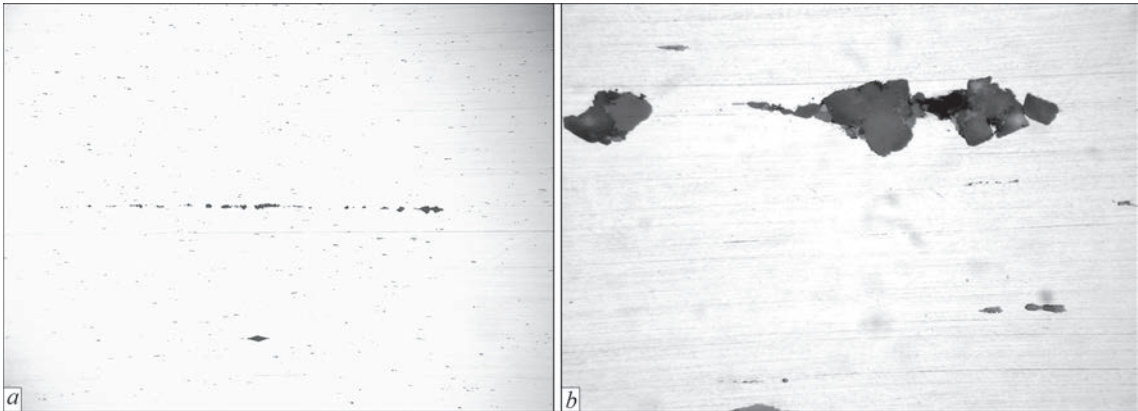


Figure 3. Complex nonmetallic inclusions in the steel $\delta = 10$ mm: *a* — $\times 100$; *b* — $\times 1000$

[11], as well as the low carbon content $C = 0.05\%$ with an oxygen content $O = 0.025\text{--}0.03\%$, steel under study can be considered semi-killed.

The phosphorus content $P = 0.05\%$ in a 12 mm thick specimen does not exceed the permissible value for VSt3sp(killed) steel [12]. The low carbon content $C \leq 0.05\%$ with a relatively moderate sulphur and phosphorus content should ensure satisfactory weldability of semi-killed, low-carbon steel.

Metallographic examinations revealed that steel specimens with a thickness of 10 and 12 mm contained mainly small light grey inclusions elongated along the rolling direction (Figures 2–3). In the 12 mm specimen, these inclusions are thinner and more elongated (Figure 2, *a, b*), while in the 10 mm specimen, they are shorter and oval in shape (Figures 2, *c, 3, a, b*).

The microhardness of these inclusions is quite high: $HV_{0.01} = 5357\text{--}6142$ MPa. Examinations using a JSM scanning electron microscope with a Link

860–500 energy dispersive spectrometer have shown that these inclusions can be classified as manganese sulphides. Isolated linear inclusions of complex composition containing manganese silicate and calcium oxide were also found (Figure 3).

Contamination with manganese sulphides according to GOST 1778–70 [9] cannot be assessed due to the small size of the inclusions. The elongated lines of complex inclusions correspond to grain size No. 2 (linear oxides and silicates) [9].

To study the microstructure of steel, 12 and 10 mm specimens were etched in a 4 % alcohol solution of nitric acid for $t = 10$ s. As a result, it was found that steel has a ferrite-pearlite structure with a small amount of cementite at the grain boundaries (Figure 4), while the volume fraction of pearlite is small. It should be noted that the microstructure of the 10 mm thick specimen is somewhat heterogeneous: along with grains of size 7–8, there are individual grains of size 4–5 [13]. The

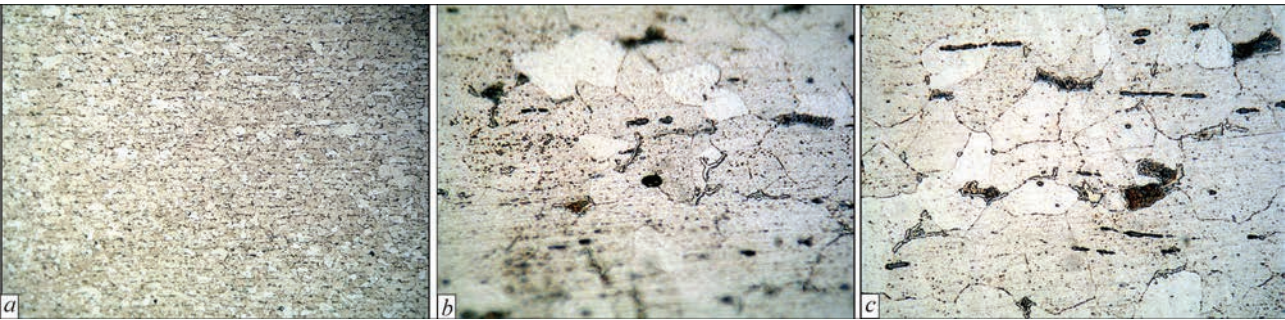


Figure 4. Microstructure of steel, $\delta = 10$ mm: *a* — $\times 100$; *b* — $\times 1000$; *c* — $\delta = 12$ mm, $\times 1000$

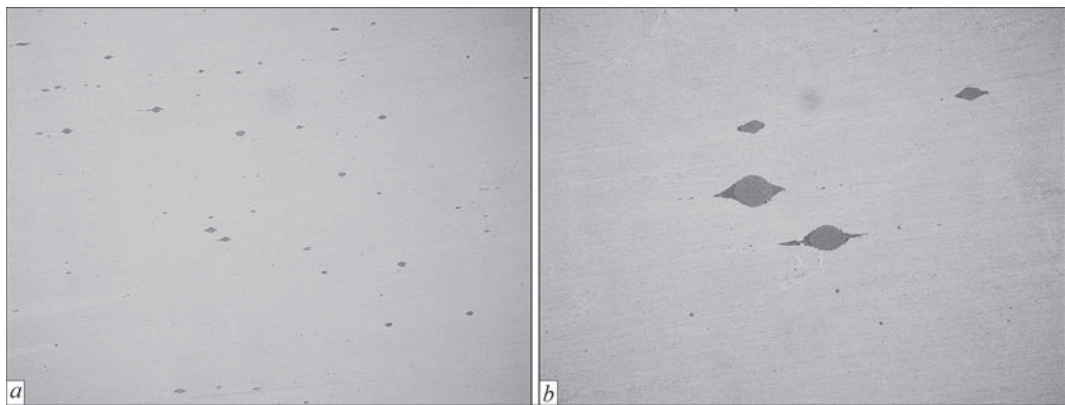


Figure 5. Sulphides and oxysulphides in steel, $\delta = 15$ mm: *a* — $\times 100$; *b* — $\times 500$

structure of the 12 mm specimen is more homogeneous and corresponds to grain size number 7–8 [13].

When measuring steel hardness according to Vickers, it was found that the hardness of a 12 mm thick metal specimen is lower than that of a 10 mm thick specimen: $HV_1 = 1163\text{--}1204$ MPa and $HV_1 = 1377\text{--}1428$ MPa, respectively.

Thus, based on the metallographic examinations, it can be concluded that, due to the fact that the metal structure is dominated by fairly fine ferrite grains and there are few dispersed nonmetallic inclusions deformed along the rolling direction and uniformly distributed between the grains, steel should be expected to have sufficiently high ductility and toughness. The actual grain size of steel and the volume fraction of pearlite should ensure the required level of steel strength. Steel is supplied in hot-rolled form.

The results of the study of the mechanical properties of a steel specimen from the tank built in 1901, compared with the required mechanical properties of VSt3sp(killed) carbon steel rolled sheet in accordance with GOST 380–60 [12], which was used for critical welded structures [14], are given in Table 3. Analysis of these data shows that, taking into account the results of testing longitudinal specimens, the strength and ductility of the tested steel meet the requirements of GOST 380–60 [12] for VSt3sp steel. The impact

toughness of the studied steel is sufficient at a test temperature of $T = -20$ °C: $KCU > 29$ J/cm² [12].

RESEARCH ON STEEL AND ITS WELDED JOINTS IN THE SHELLS OF THE TANK BUILT IN 1929

The research was conducted for the course I of the tank shell with a thickness of 15 mm. The chemical composition of steel is shown in Table 4. It can be seen that its grade is similar to that used for the shell of the tank built in 1901, but the sulphur content significantly exceeds the permissible values specified in GOST 380–60 [12] by a factor of two, while the phosphorus content meets the requirements.

The study of nonmetallic inclusions showed that the metal mainly contains light grey sulphides of a spindle-shaped form, elongated in the direction of rolling (Figure 5). The length of these inclusions is 10–75 μ m. The sulphide contamination size is 5b [9]. The base metal also contains small amounts of oxysulphides and lamellar silicates (Figure 6, *a*), and sometimes chains of oxides oriented in the direction of rolling (Figure 6, *b*).

It was found that the structure of the base metal is ferrite-pearlite (Figure 7, *a*, *b*), with pearlite elongated along the rolling direction. A small amount of cementite was also observed (Figure 7, *c*), etching in an alkaline solution of sodium picrate. The diameter

Table 3. Mechanical properties of the steel used for the shells of the tank built in 1901

Structural element	Mechanical properties of steel				
	σ_y , MPa	σ_p , MPa	δ_5 , %	KCU_0 , J/cm ²	KCU_{-20} , J/cm ²
Course I, $\delta = 12$ mm*	278.4 273.1	386.2 381.9	37.3 40.0	—	74 69 32
Course II, $\delta = 10$ mm*	—	—	—	122 120 118	—
Steel VSt3sp** [12]	> 240	380–400	> 27	—	> 29**

*Longitudinal specimens. ** Optional requirements.

Table 4. Chemical composition of the steel in the shell of the tank built in 1929

Metal thickness δ , mm	Content of chemical elements, mass fraction, %						
	C	Si	Mn	S	P	N	O
15	0.045	< 0.05	0.43	0.10	0.05	0.0026	0.027

of ferrite grain is $d = 80\text{--}90\text{ }\mu\text{m}$. The hardness of the base metal is $HV_1 = 1560\text{ MPa}$.

A study of the structure of a welded butt joint shows that ferrite with finely dispersed pearlite forms in the overheated area (Figure 8, *a*). Small amounts of cementite are observed at the grain boundaries. The ferrite grain size increases insignificantly and is $80\text{--}90\text{ }\mu\text{m}$. The hardness of the metal in the overheated area is $HV_1 = 1917\text{ MPa}$. Near the fusion line, sulphide inclusions under the influence of welding heat

have a shape closer to oval. The maximum length of sulphides is $15\text{--}20\text{ }\mu\text{m}$ (Figure 6, *c*). In the normalized region, steel structure is slightly refined, and its hardness is $HV_1 = 1800\text{ MPa}$. During partial recrystallization, some refinement of ferrite, spheroidization and refinement of pearlite, and formation of a small amount of cementite occur (Figure 8, *b*). The hardness of this region is $HV_1 = 1725\text{ MPa}$.

Analysis of the results of the microstructure examination of the welded joint shows that an increase



Figure 6. Lamellar silicates in steel: *a* — $\delta = 15\text{ mm}$, $\times 500$; *b* — chain of oxides in steel, $\delta = 15\text{ mm}$, $\times 500$; *c* — sulphides and oxysulphides near the fusion line, $\delta = 15\text{ mm}$, $\times 500$

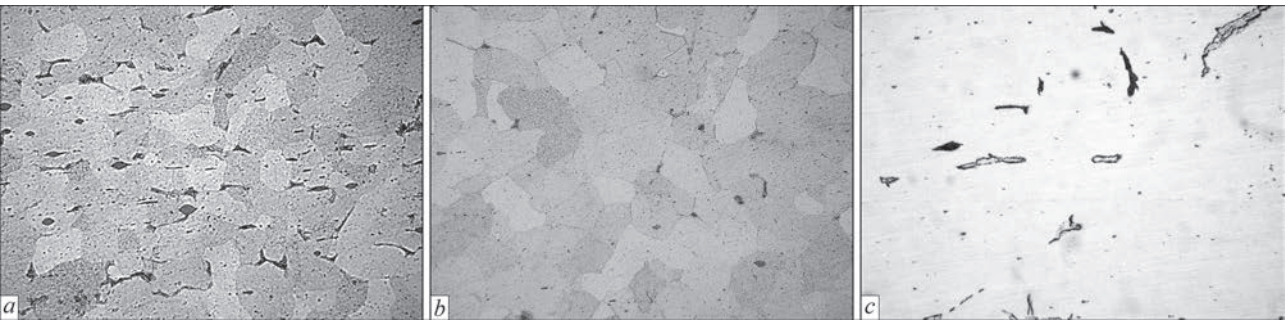


Figure 7. Microstructure of the base metal, etching in nitric acid: *a* — $\times 250$; *b* — $\times 500$; *c* — etching in an alkaline solution of sodium picrate, $\times 250$

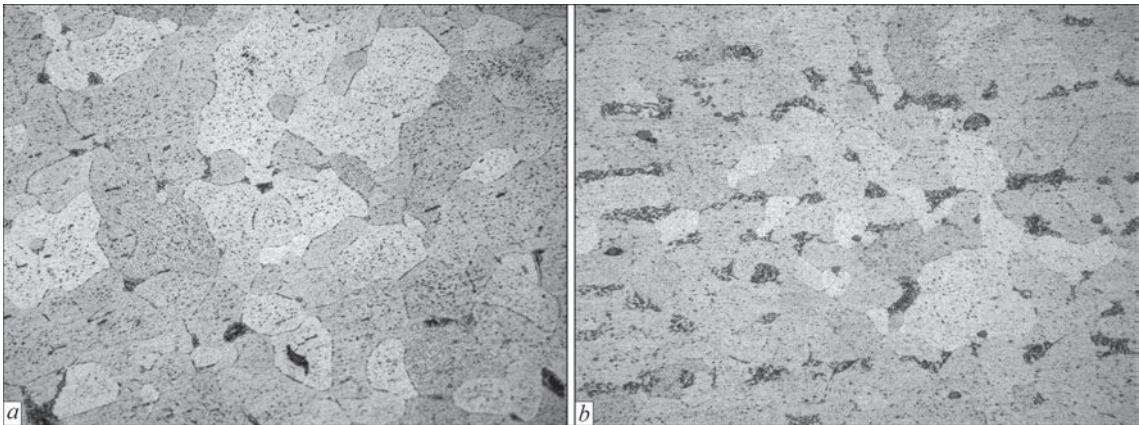


Figure 8. Microstructure of HAZ, $\times 250$: *a* — coarse grain area; *b* — partial recrystallization area

Table 5. Results of tensile tests of the steel used for the shells of the tank built in 1929

Structural element	Mechanical properties of steel		
	σ_y , MPa	σ_t , MPa	δ_5 , %
Girth I, $\delta = 15.0$ mm*	189	335	38.7
	186	336	40.3
Steel 08ps(semi-killed) [11]	> 196	> 320	> 33

*Transverse specimens.

in hardness in the overheated area is associated with the formation of a coarser finer-dispersed pearlite in larger volume, which is uniformly distributed over the grain area (Figures 7, *a*, 8, *a*). In the normalized and partially recrystallized area, an increase in hardness occurs mainly due to the formation of a finer-grained structure during cooling (Figure 8, *b*).

In general, analysis of the obtained data shows that in the welded joint of the studied specimen of steel from the tank built in 1929, there is no significant growth of ferrite grains in the coarse grain area; instead, a sufficiently fine-grained ferrite-pearlite structure with a small volume fraction of pearlite is formed. Precipitates on the grain boundaries are insignificant. There are no elongated nonmetallic inclusions in the base metal and near the fusion line. This gives grounds to expect a sufficiently high level of metal viscosity in the coarse grain area and the base metal.

The results of determining the mechanical properties of the studied steel during tensile testing of specimens are presented in Table 5. Analysis of the test results shows that the strength of steel practically meets the requirements for steel 08ps(semi-killed) GOST 1050–88 [11].

In the second stage of research, the impact toughness of steel and HAZ was determined on the specimens with a V-shaped notch with a cross-section of 10×10 mm. In [6], it is stated that based on the processing of statistical data on the mechanical properties of various steel grades used for ships that underwent brittle fracture, the main criterion for assessing the probability of brittle fracture was the temperature at which the fracture energy according to the Charpy method was equal to $KV = 2$ J ($KCV = 24$ J/cm²). It was established that in the case of welded ships for the studied steel grades (carbon, rimmed type with relatively low impact toughness), the specified test temperature for the specimens is $T = +15.5$ °C (60 F) [6]. For Batumi, the temperature of the coldest days with a probability of 98 % is $T_{0.98} = -4$ °C [15], and for Baku, $T_{0.98} = -8$ °C [15]. In these cases, the calculated metal temperature is $T = +1$ °C (33.8 F) and $T = -3$ °C (26.6 F), respectively [16], which is sufficient to exclude brittle shell fracture, taking into account the

Table 6. Results of impact toughness tests on the base metal and fusion line of the steel used for the shell of a riveted tank built in 1929

Tank characteristics	Specimen testing temperature, °C	Impact toughness KCV , J/cm ²	
		Of base metal*	Notch along the fusion line**
t15 mm, No. 1	+15.5 (60 F)	<u>24–28–20</u> 24	–
	0 (32 F)	<u>23–20–20</u> 21	<u>158–129–125</u> 137
t15 mm, No. 2	0 (32 F)	<u>30–20–23</u> 24	–

*Specimens for Charpy impact testing were cut out across the rolled material.

**The location of the V-shaped notch was taken along the fusion line at pos. 2 [18].

thickness of the courses [17]. Therefore, considering the climatic region where the tanks are located, the application of the above criterion is quite correct.

Taking into account the heterogeneity of the mechanical properties of sheet metal, and also to assess the reliability of the lower limit of the transition interval, the specimens were tested at two temperatures: $T = +15.5$ °C (60 F) and $T = 0$ °C (32 F). The obtained value of the impact toughness of the base metal and the fusion line are given in Table 6.

Analysis of the data in Table 6 shows that the base metal has sufficient toughness to avoid brittle fracture when welding is used – the toughness level is $KCV_{+15.5} = 24$ J/cm² and $KCV_0 = 21$ J/cm² [6, 17]. The fractures of the specimens at a test temperature of $T = 0$ °C are brittle, having a shrinkage depression of the specimens in the lower part being ≈ 0.34 mm (Figure 9). The high impact toughness of the specimens with a notch along the fusion line (FL) is associated with the presence of a high proportion of weld metal, which has a significantly higher toughness in the fracture cross-section. Visual examination of the

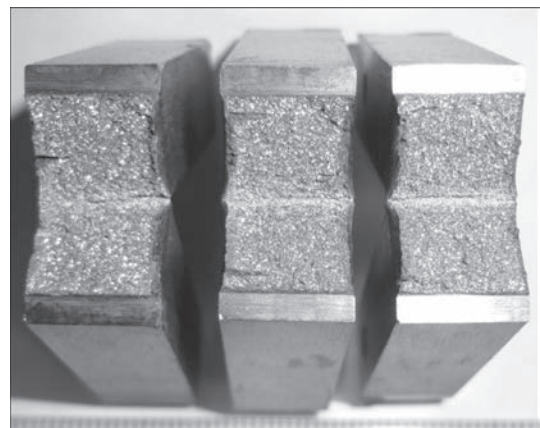


Figure 9. Fractures of specimens tested according to the Charpy method at $T = 0$ °C (course I, thickness $\delta = 15$ mm, tank built in 1929)

specimen fractures shows that there is no transition of the FL metal into brittle state. This means that when assessing the impact toughness of a welded joint, considering the existing structure of the FL metal, one can refer to the toughness indices of the base metal.

To assess the adequacy of steel fracture toughness to prevent brittle fracture of the tank, the critical crack length for the shell course I with a thickness of 15 mm was determined. Based on the impact toughness values KCV_0 , the critical fracture toughness of steel was determined using the formula [19]:

$$K_{Ic} = \sqrt{\frac{0,1 \cdot E \cdot KCV}{10^2(1-\nu^2)}} \quad (1)$$

where K_{Ic} is the critical intensity coefficient for stress conditions under which crack edges shift in the direction normal to the crack plane, $\text{MPa} \cdot \sqrt{\text{m}}$; E is the Young's modulus, for steel $E = 210,000 \text{ MPa}$; KCV is the impact toughness value, $KCV = 20 \text{ J/cm}^2$; ν is the Poisson's ratio, for steel $\nu = 0.25$.

In this case, the critical crack length is [20]:

$$a_{cr} = \frac{\left(\frac{K_{Ic}}{\sigma \cdot F}\right)^2}{\pi} \quad (2)$$

where σ is the stress acting in the direction normal to the crack plane in the course I of the shell, MPa . Taking into account the ratio $2l/b = 0$ ($2l$ is the crack length; b is the plate width), the value of the function F was taken as $F(0) = 1$ [20].

The stresses σ acting in the direction normal to the crack plane were determined according to [17] $\sigma = 2/3 \cdot Y = 2/3 \cdot 186 \text{ MPa} = 124 \text{ MPa}$, where Y is the minimum yield strength value taken according to Table 5.

The results of calculating the critical length of a through crack for the base metal of the course I, obtained on the basis of its impact toughness values (Table 6), are given in Table 7.

Table 7 shows that the critical length of a through crack a_{cr} is 97 mm. The initial length of a through crack from which oil leakage begins can be taken as in the case of a semi-circular surface cavity: $2t$, where t is the shell thickness or $2 \times 15 \text{ mm} = 30 \text{ mm}$ [21]. In this case, the crack should increase by 67 mm before reaching the critical length. The expected number of cycles to achieve a through crack of critical length can

be conservatively estimated as $N = 1000$ [22]. In this case, with a margin of $N = 10$ [23], the expected number of cycles for a through crack is $N = 100$, which corresponds to one year of tank operation. During this period, a through defect will be guaranteed to be detected during regular external inspection of the tank shell, which allows using the "leak before failure" criterion to ensure trouble-free operation of the tanks [21–24].

Thus, a distinctive feature of the investigated steel grade is its low carbon content (0.045–0.05 % C) combined with partial deoxidation of steel (semi-killed) and an increased sulphur content of up to 0.1 % S and phosphorus content of up to 0.05 % P. This means that the cold resistance of steel will be determined by a ferritic matrix, which has a sufficient level of ductility and toughness. In turn, the fracture energy of the metal is also influenced by the number and size of nonmetallic inclusions [25]. In [26], it is shown that for C–Si–Mn steels, an increase in the sulphur content to 0.1 % does not significantly affect the lower limit T_0 of the critical brittleness interval. According to the obtained data (Table 6), the impact toughness of steel is $KCV_0 = 20 \text{ J/cm}^2$, which provides a sufficient lower limit of the transition temperature T_0 for climatic regions where tanks are built [17].

In the case of welding the studied low-carbon steels, there is no brittle fracture of the metal in the fusion line, which is associated with the formation of a predominantly ferritic structure (low carbon content $C \leq 0.05 \%$), an insignificant increase in the size of the ferrite grain and the absence of brittle phase precipitation at the grain boundaries in the metal. All this ensures the exclusion of brittle fracture of welded joints in the studied steels and provides a basis for the repair of riveted tanks using welding.

KEY TECHNICAL DECISIONS MADE DURING THE REPAIR OF RIVETED TANKS

Tank built in 1901, $D = 23.40 \text{ m}$, $H = 9.54 \text{ m}$. An inspection of the technical condition of the riveted tank revealed the need to repair the base, completely replace the bottom and conical roof of the tank, and install new pipes and hatches in the shell and roof.

To reinforce the lower part of the shell, which had suffered from pitting corrosion, a 6 mm thick bandage made of S235JR steel was welded inside the tank (Figure 10). The base was restored and a new bottom was installed: 8 mm thick annular bottom plates and a 6 mm thick central part made of S235JR steel. The outer contour sheets of the bottom remained. New DN600 manholes and new pipes were cut out into the shell (Figure 10). A new conical frame roof with a 6° angle of inclination was designed with a central

Table 7. Results of calculating the critical crack length for the course I of the riveted tank built in 1929, base metal

Impact toughness KCV_0 , J/cm^2	Stress intensity factor, K_{Ic} , $\text{MPa} \cdot \sqrt{\text{m}}$	Tensile stress σ , MPa	Critical crack length a_{cr} , mm
20	67	124	97

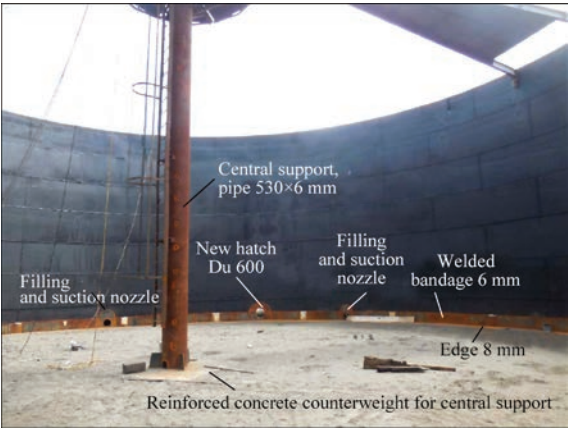


Figure 10. Repair process of the riveted tank, 1901 year, internal view

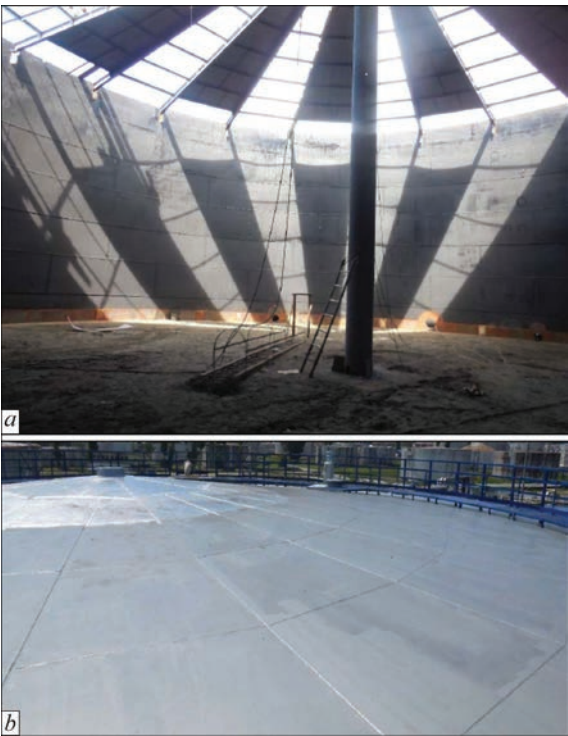


Figure 11. Repair of the conical roof of the riveted tank, 1901 year: *a* — mounting of roof panels, view from the inside; *b* — mounted roof, view from the outside

DN500 support, which was attached to a reinforced concrete load to prevent it from rising under the ac-



Figure 12. Repair stringers installed at the points where the radial beams are connected to the shell



Figure 13. Dismantling of the existing bottom and roof of the tank, 1929 year

tion of excess pressure in the tank (Figure 10). The main radial beam was an I-beam No. 27, and new roof plates with thickness 4 mm. The I-beams rested on the shell through support tables, and the roof was installed using factory shields, with a factory set and roof plates installed in bulk between the shields (Figure 11, *a, b*). On the outside of the tank, the place where the radial I-beams rested on the shell was additionally reinforced with stringers in the form of a 100×100×8 mm equal-sided angle (Figure 12).

To reduce metal heating during welding and obtain a favourable structure and mechanical properties of the HAZ metal, multi-pass welding using coated basic electrodes with a diameter of 3.2 mm was used.

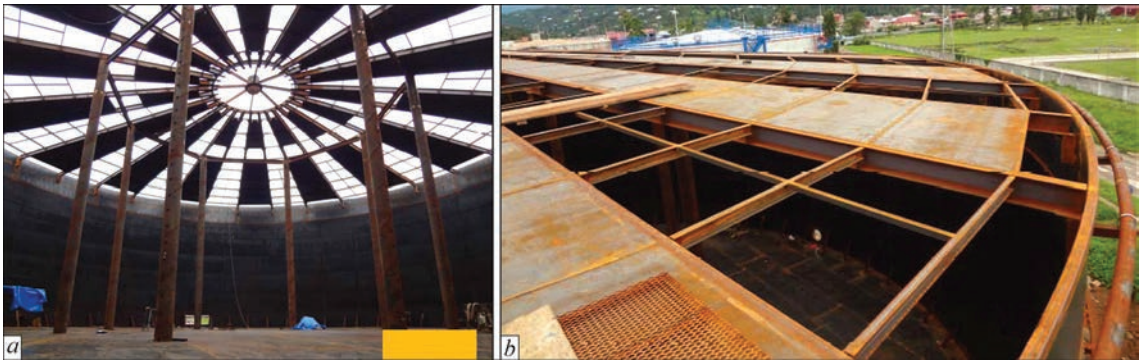


Figure 14. Installation of a conical roof with support in the centre: view from inside the tank, 1929 (*a*), view from outside the tank (*b*)



Figure 15. Installation of a reinforced concrete circular foundation for the supporting columns of the roof



Figure 16. Installation of a new bottom: annular bottom plates of 8 mm thick, central part of 6 mm thick



Figure 17. Replacement of the upper shell course with mounting of a fixing ring



Figure 19. View of the tank, 1929 year, after repair during the application of an anti-corrosion coating

After repairs using welding, the tank's operability was restored, achieving the maximum possible fill level.

Tank built in 1929, $D = 30.73$ m, $H = 10.25$ m. During the repair of this tank, the base was repaired, the bottom, roof and upper shell course were replaced (Figure 13). The slope of the new conical frame roof was 1:10. An I-beam No. 27 was used as the main radial beam, which was supported on the shell by a 0.5 m long support table. The new conical roof was supported in the centre by a beam ring made of an I-beam No. 27, which in turn was supported by a frame of beams on 8 columns made of $\varnothing 426 \times 8$ mm pipe, which were installed on a circular reinforced concrete foundation with a radius of $R = 7$ m (Figures 14, a, b, 15). The thickness of the new roof plates was 4 mm. During the restoration of the bottom, its central part was dismantled, leaving a segment of the edge around the perimeter adjacent to the shell of 600 mm long. The base was restored and a new bottom was installed: an 8 mm thick annular bottom plates, a 6 mm thick central part made of S235JR steel (Figure 16). As a result of intense pitting corrosion, the upper seventh shell course of 1400 mm high was replaced with a new one of 6 mm thick, also made of S235JR steel. In order to preserve the design geometric shape of the upper shell courses, a stiffening ring made of square-section pipe of $80 \times 80 \times 4$ mm was preliminary installed on the sixth course (Figure 17). For

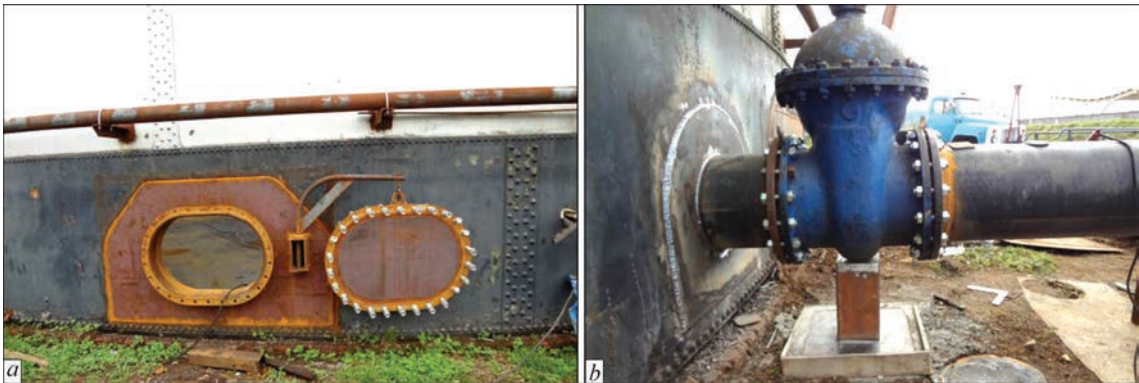


Figure 18. Welding of an oval manhole (a) and filling and suction nozzle (b) into the shell

ease in service, new oval manholes 600×900 mm and round DN600 manholes were installed in the shell, and the nozzles were replaced with new filling and suction nozzles (Figure 18, *a, b*).

The above-mentioned repair to the tank made it possible to restore the design oil filling level (Figure 19).

For welding butt joints of the shell, process 135 was used in a mixture of M21–ArC–18 using G3Si1 wire with a diameter of 1.2 mm. During the repair, special attention was paid to non-destructive testing of the repaired butt welds in the course I — all repaired butt welds were tested along their entire length using an ultrasonic method in accordance with the requirements of SNiP 3.03.01–87 [27]. During visual and measurement inspection of the repair welds, the depth of undercuts was allowed to be not more than 0.5 mm. Welders were permitted to perform welding works based on the positive results of the qualification test joints. The combination of the above organisational and technical measures ensured high quality welds and guaranteed the absence of dangerous elongated crack-like defects, such as lack of fusion penetration at the edge and root of the weld, slag inclusions, and undercuts, which can be the site of brittle fracture initiation.

After successful hydraulic testing, the tanks were put into operation. Long-term operation of the restored tanks (over 10 years) has shown that the technical solutions developed on the basis of research into the possibility of using welding to repair riveted tanks can ensure their sufficiently reliable further operation with the maximum permissible level of petroleum product filling.

CONCLUSIONS

Steel used in the construction of the investigated riveted tanks is low-carbon steel with a carbon content of 0.05 % C and manganese content of 0.4–0.5 % Mn, which can be classified as structural semi-killed steel.

Although the sulphur content in steel can reach 0.1 % S, given the existing structure of steel and its HAZ, which mainly consists of ferrite, and the absence of elongated low-melting sulphides and phase precipitates at grain boundaries, the cold resistance of the studied steel and its welded joints satisfies the condition for excluding brittle fracture in the case of welded steel structures: impact energy $KV \geq 20$ J ($KCV \geq 24$ J/cm²) at a specimen test temperature $T = +15.5$ °C (60 F), which corresponds to the climatic region where the tanks are located.

The calculated minimum defect size for the impact toughness of steel $KCV_0 = 20$ J/cm² is sufficient for the application in repair butt welded joints of shells

according to the “leakage to failure” criterion, which allows using welding in the repair of riveted tanks and ensures their subsequent long-term trouble-free operation after repair.

To determine the possibility of repairing riveted tanks using welding, it is necessary to conduct a comprehensive study of the physical and mechanical properties of steel and its repair welded joints by specialized organizations. The welding technology must ensure minimum heat input, which minimizes grain growth and prevents the formation of phases at grain boundaries.

For repair butt welded joints in the shells, ultrasonic quality testing must be performed along their entire length. Particular attention should be paid to detecting elongated plane defects along the edge or at the root of the weld.

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CONFLICT OF INTEREST

The Authors declare no conflict of interest

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INFLUENCE OF DYNAMIC LIQUID MOLTEN LAYER ON JOINT FORMATION DURING MIAB WELDING OF PIPES

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ABSTRACT

The results of experimental and theoretical studies of the process of magnetically impelled arc butt (MIAB) welding of thick-walled pipes made of carbon and high-strength pipeline steels are presented. The study shows that the key factor in the formation of a high-quality welded joint is the controlled formation and retention of a continuous layer of molten metal in the gap between the pipe ends during the period immediately preceding upsetting. It is shown that the controlled motion of the arc in an external magnetic field ensures uniform heating along the perimeter and through the thickness of the edges, reduces local overheating, and minimizes deviations from the optimal values of the temperature field and heat-affected zone (HAZ) microstructure. A regime of combined arc-electrical contact heating is identified, in which large-area liquid contacts maintain a continuous molten layer even when welding pipes with a wall thickness exceeding the diameter of the active arc spots. It is experimentally confirmed that welded joints of pipes with diameters up to 320 mm and wall thicknesses up to 16 mm made of steels X70, X80, 09G2S, 35, STPG410, and 01Star520 exhibit strength and ductility characteristics at the level of the base metal, while the impact toughness at temperatures down to $-40\text{ }^{\circ}\text{C}$ is at least 80 % of that of the base metal. The process is implemented without the use of shielding gases and without subsequent post-weld heat treatment.

KEYWORDS: magnetically impelled arc butt (MIAB) welding of thick-walled pipes, controlled magnetic field, molten metal, heat-affected zone, impact toughness

INTRODUCTION

Ensuring stable quality of welded joints in non-rotatable butt joints of steel pipes with high productivity and minimal influence of the human factor is one of the key challenges for enterprises in the energy, oil and gas, and mechanical engineering sectors. This problem is particularly critical when welding thick-walled pipes and tubular components with wall thicknesses exceeding 5 mm made of high-strength steels, where even slight deviations in the thermal regime and deformation conditions lead to significant variability in the structure and mechanical properties of welded joints.

MIAB welding belongs to high-productivity pressure welding processes and combines concentrated arc heating with controlled displacement of the heat source in an external magnetic field [1–7]. The industries in which the MIAB process is already applied include the construction of pipelines for soil thermal stabilization systems at ambient temperatures down to $-40\text{ }^{\circ}\text{C}$ — more than 8,000 km of pipelines with diameters up to 100 mm have been welded, maintaining a total of 25 million cubic meters of soil in a frozen state over an area of 2.5 million square meters; the construction of greenhouse complexes — more than 100 hectares of small-diameter pipes up to 70 mm in diameter have been welded, including pipes with fit-

tings for various applications; the production of drive-shafts in the automotive industry; and the manufacture of tubular housings for hydraulic cylinders and other components [8–10].

THE AIM OF THIS WORK

is to conduct a study followed by experimental validation of the influence of the dynamic liquid molten layer during the welding process on joint formation, as well as to develop a technology of MIAB welding of pipes and tubular components with diameters up to 320 mm and wall thicknesses up to 16 mm made of high-strength steels, adapted for both field and stationary operating conditions.

PHYSICAL PRINCIPLES OF THE MIAB PROCESS

The essence of the welding process (Figure 1) consists in heating the ends of the welded pipes by an arc discharge moving within the gap under the action of an external magnetic field, followed by upsetting of the heated ends and solid-phase joint formation after expulsion of molten metal [11–15].

The arc motion is determined by the action of electrodynamic forces arising from the interaction between the arc current I and the magnetic field induction B . The radial and axial components of the magnetic field together with the welding current generate forces that ensure both the rotational motion of

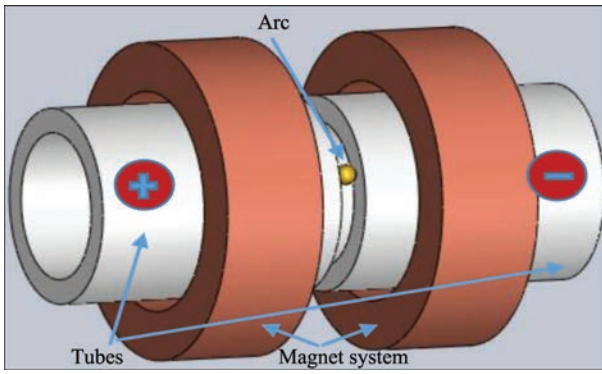


Figure 1. Schematic illustration of the MIAB pipe welding process

the arc along the pipe perimeter and its transverse oscillations between the inner and outer surfaces of the edges. An important role in this process is played by the ferromagnetic mass of the pipes, which affects the distribution of the magnetic field and promotes a stable displacement of the arc, after ignition, toward the inner edge region of the pipe ends [16].

The intensity of heating of the pipe ends is determined by the arc current, the arc travel speed, and the scanning pattern. Unlike a stationary arc, a rapidly moving arc under the action of an external magnetic field forms an averaged temperature field with a reduced maximum temperature in the near-contact layers, which is of fundamental importance in press welding high-strength pipeline steels.

INFLUENCE OF THE MOLTEN LAYER ON WELDED JOINT FORMATION

As with other pressure welding methods, obtaining high-quality joints during MIAB welding requires the establishment of a prescribed temperature field and protection of the heated pipe ends. It is considered [16] that partial protection against oxidation in the gap is provided by metal vapors generated during arc heating.

The technologies developed at the PWI are based on the concept that, similar to flash-butt welding [17], the main condition for obtaining high-quality joints in MIAB welding is the formation, on the ends of the welded components during the period preceding upsetting, of a layer of molten metal and oxides of the alloying elements contained therein. Due to the heating by the rapidly moving arc, the molten layer reaches

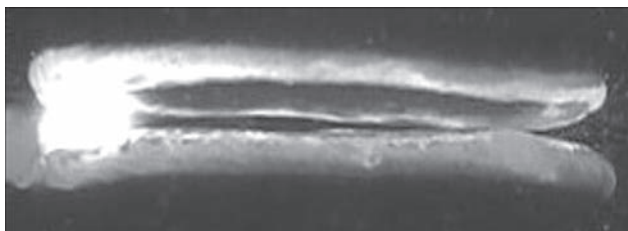


Figure 2. Molten layer formation on pipe ends due to arc motion

a temperature above the melting point of the welded material. Experimental studies have shown that, during contact melting of steels, the melt temperature reaches up to 1700 °C, and its solidification occurs in the range of 1700–1370 °C. If deformation of the heated pipe ends is performed within this temperature range, the molten metal is completely expelled from the joint along with oxide structures. The duration of the melt solidification interval, t_l , depends on the thermophysical properties of the welded materials and the adopted welding technology.

$$t_l = \frac{\delta_l \gamma q}{\lambda \frac{d\theta}{dx}_{x=0}}, \quad (1)$$

where t_l is the duration of solidification; δ_l is the thickness of the molten layer; $\frac{d\theta}{dx}_{x=0}$ is the temperature gradient in the near-contact layer; λ is the thermal conductivity; q is the heat content of the molten metal at temperatures above its melting point.

During the selection of the welding technology, it is possible to vary the heating conditions of the pipes and, to some extent, the thickness of the molten layer, δ_l . The most favorable conditions for obtaining a high-quality welded joint are achieved by ensuring a sufficient molten layer thickness and an optimal temperature gradient in the near-contact layer, which allows the deformation of the pipe ends to be carried out within the solidification range. Under such conditions, the molten metal along with oxide inclusions is effectively expelled from the joint, providing cleaning of the weld zone and formation of a homogeneous fine-grained structure with enhanced mechanical properties.

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Experimental studies using high-speed video recording have shown that during heating, a molten layer forms on the pipe ends, and moves in the same direction as the welding arc (Figure 2). The thickness

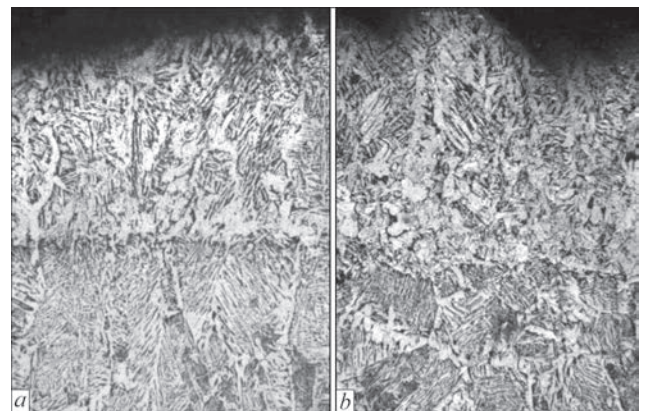


Figure 3. Microstructure of the cast metal at the cathode (a) and anode (b) (×200)

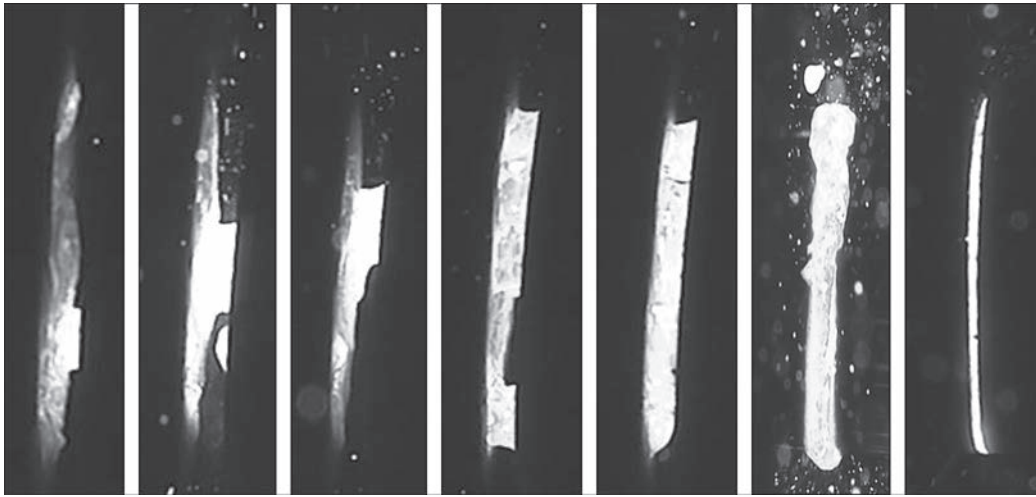


Figure 4. Sequence of frames showing the formation, evolution, and collapse of the dynamic molten layer between pipe ends before and during upsetting

of this layer during welding can reach up to ~ 0.6 mm (Figure 3).

As the temperature increases, the molten metal is retained on the pipe-end surfaces by surface tension forces, and upon reaching a critical thickness, it begins to move synchronously with the arc.

The motion of the molten layer is caused by the combined action of electrodynamic forces and gas-dynamic pressure in the arc gap. This motion leads to intensive mixing of the molten metal, its renewal, and leveling along the perimeter and through the thickness of the pipe edges.

Using high-speed video recording (5000 frames/s), the formation of the molten layer on the pipe ends during heating by a moving arc was studied in the interval immediately preceding upsetting.

In Figure 4, a series of high-speed frames illustrates the process of formation of a dynamic molten metal layer in the arc gap between the pipes under arc heating conditions. It can be seen that the melt does not form a stationary weld pool, but exists as a thin, elongated, and unstable layer that continuously deforms, stretches along the gap axis, and locally breaks under the action of thermal, electromagnetic, and hydrodynamic forces.

Under the action of an external magnetic field, the liquid contacts undergo controlled motion over the entire area of the pipe ends, which ensures the preservation of a continuous molten layer up to the moment of upsetting.

During the evolution of the dynamic layer, its expansion and an increase in the area of liquid contact between the ends are observed, which leads to gradual extinction of the arc discharge and transition of the process to the mode of electrical contact conductive heating. In the final frames, the collapse of the mol-

ten layer and the establishment of a continuous liquid contact prior to upsetting are recorded.

Beneath the molten layer, solid metal heated to a plastic state is located. After the upsetting force is applied, direct contact and joint plastic deformation of the material occur, which creates the conditions for the formation of a continuous welded joint.

The results obtained indicate that the state and dynamics of the molten layer in the interval immediately preceding upsetting are the determining factors in forming high-quality joints during MIAB welding. The dynamic molten layer performs a multifunctional role: it shields the pipe ends from oxidation, promotes the disruption of oxide films, and facilitates the removal of oxidation products and alloying impurities from the contact zone.

Premature solidification of the molten layer before deformation leads to the formation of oxide inclusions in the joint plane and a reduction in the mechanical properties of the weld.

Implementation of such thermal and electrodynamic conditions is a sufficient criterion for ensuring stable welding quality.

One of the key outcomes of this study is establishing the correlation between the uniformity of the molten layer and the reduction of variability in weld properties. During MIAB welding, uniform distribution of the molten metal along the pipe perimeter results in:

- leveling of the temperature field in the near-contact zone;
- reduction of deviations in the HAZ width;
- formation of a homogeneous fine-grained structure along the joint line;
- reduction of microhardness variability and improvement of impact toughness (KCV) of the welded joints.

Figure 5 shows the temperature field distribution in the heating zone during MIAB welding of high-strength steel pipes. The temperature of heating of the near-contact metal layers at a distance of 2 mm from the melting surface immediately before the onset of deformation does not exceed 1050 °C. Experimental studies have shown that under rapid heating of steel pipe specimens to temperatures above 1100–1150 °C, a sharp decrease in Charpy impact toughness (KCV) is observed, which is associated with intensified grain growth and formation of structural inhomogeneities.

Since, during upsetting, the microstructure of the central part of the weld and the joint line zone is determined by the structural state of the near-contact layers, this temperature distribution can be considered optimal for various pressure welding methods in terms of achieving high KCV values close to those of the base metal and enhanced resistance of welded joints to brittle fracture.

Despite the symmetric geometry of the process, the temperature in the cathode zone slightly exceeds that of the anode, which is caused by differences in the energy transfer mechanisms in the near-contact layers of the arc. At the cathode, the main contribution to the heat balance is provided by bombardment of the surface with positive ions accelerated in the cathode voltage drop, whereas at the anode, heating is primarily due to electrons with lower energy. The higher cathode voltage drop (10–20 V) compared to the anode (2–5 V) results in a significantly higher heat flux density at the cathode, leading to a higher local cathode temperature compared to the anode [18–20].

During the MIAB welding process, the stable molten layer enables upsetting of the pipes within the optimal solidification temperature range. This ensures effective expulsion of the liquid phase and oxide inclusions, resulting in the formation of a joint in the solid phase with reduced sensitivity to local overheating.

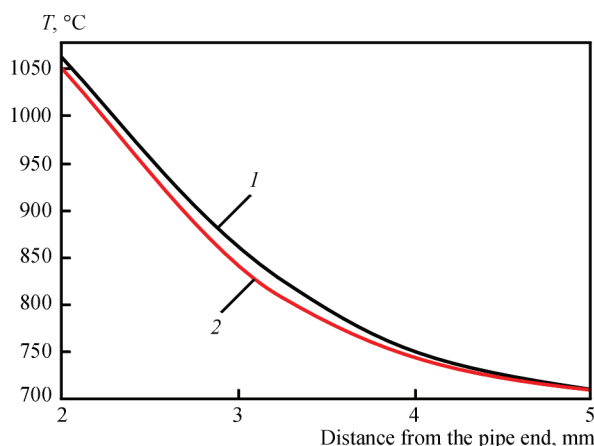


Figure 5. Formation of the temperature field during pipe welding: 1 — cathode; 2 — anode

EXPERIMENTAL STUDIES AND RESULTS

Welding was performed on MIAB welding machines of types MD-101 for pipes with diameters up to 57 mm (Figure 6) and MD-205 for pipes with diameter up to 220 mm (Figure 7).

The studies were conducted on pipes made of steels X70, X80, 09G2S, 35, STPG410, and 01Star520, with diameters ranging from 42 to 320 mm and wall thicknesses of 4–16 mm. Process control included registration of current, voltage, arc travel speed, and upsetting force.

Mechanical testing of welded joints, performed according to the standards of the American Petroleum Institute (API) [21] and the American Welding Society (AWS) [22], showed that the tensile strength and ductility properties are at the level of the base metal, while the Charpy impact toughness (KCV) at +20, –20, and –40 °C is not less than 80 % of the corresponding base metal values [23]. Importantly, these properties are achieved without any post-weld heat treatment.

Metallographic studies confirmed the formation of a narrow HAZ (approximately 4–10 mm, depending on wall thickness) and a fine-grained structure in the central part of the weld joint.



Figure 6. MIAB welding machine MD-101



Figure 7. MIAB welding machine MD-205

ANALYSIS OF PROCESS PARAMETERS' INFLUENCE ON JOINT FORMATION.

EFFECT OF ARC TRAVEL SPEED

The arc travel speed in the gap between pipe ends is one of the key parameters of MIAB welding. Increasing the arc speed leads to a more uniform distribution of heat along the perimeter and through the thickness of the edges, reducing local overheating and the likelihood of coarse-grained structures.

Experiments showed that for pipes with diameters of 42–320 mm, the optimal arc travel speed during heating is 50–240 mm/s. Within this range, a stable molten layer of the required thickness forms, and its distribution along the pipe-end perimeter is uniform.

EFFECT OF MAGNETIC FIELD INDUCTION

The induction of the external controlling magnetic field determines the nature of the arc motion and the thermal penetration depth of the pipe edges. Insufficient magnetic field induction leads to unstable arc movement, uneven heating, and increased variability in welded joint structure. Increasing the field induction to optimal values ensures controlled rotational motion of the arc, providing a uniform thermal regime.

However, excessive magnetic field induction may result in excessive arc acceleration and poor molten layer formation. Therefore, magnetic field optimization must consider pipe diameter, wall thickness, and steel composition.

EFFECT OF UPSETTING FORCE

The upsetting force during MIAB welding is a key technological parameter that determines the completeness of expulsion of the liquid phase and oxide inclusions from the joint, as well as formation of continuous metal contact over the entire joint area. Insufficient upsetting force allows part of the molten metal and oxide products to remain in the joint, causing nonmetallic inclusions and reduced mechanical properties.

Increasing the upsetting force to optimal levels ensures intensive plastic deformation of the near-contact layers, effective expulsion of the molten phase beyond the joint, and disruption of oxide films. This promotes fine-grained structure formation in the joint zone and reduces variability of mechanical properties along the weld perimeter.

At the same time, excessive upsetting force may cause over-extrusion of metal, increased weld reinforcement, and local structural inhomogeneities. Therefore, selection of the upsetting force in MIAB welding must consider the temperature state of the near-contact layers, wall thickness, and conditions for forming the molten layer before deformation.

FORMATION OF LOW-PROFILE WELD REINFORCEMENT OF THE PIPE BY MIAB WELDING

During pipe welding, the weld forms on both the external and internal surfaces of the joint. Depending on the operating conditions of the welded pipe structures, this reinforcement can be mechanically removed (Figure 8) or retained as part of the product. However, in many industrial and construction applications, significant external and internal reinforcement limits the use of pressure welding. These limitations are due to strict requirements for hydraulic and aerodynamic characteristics of pipelines and tubular components, as well as the precision of joint coaxiality in assembly structures.

Reducing the weld reinforcement height is of fundamental importance for expanding the application scope of pressure welding. The conducted studies showed (Figure 9) that by controlling the energy supplied to the pipe ends during MIAB welding process, the allowance for pipe upsetting, and the action of the external magnetic field during heating and upsetting, it is possible to purposefully influence the geometry of the welded joint.

It was established that, during rotation of the welding arc under the influence of an external magnetic field at speeds of up to 200 mm/s, a thin layer of molten metal forms on the pipe ends with a thickness of up to 0.6 mm. Due to the rotational motion of the arc,

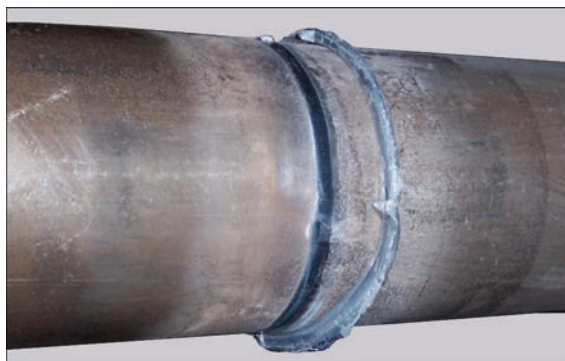


Figure 8. Pipe with dimensions of 121×10 mm, steel 01Star520, after flash removal



Figure 9. Technologically reduced weld reinforcement height on the pipe

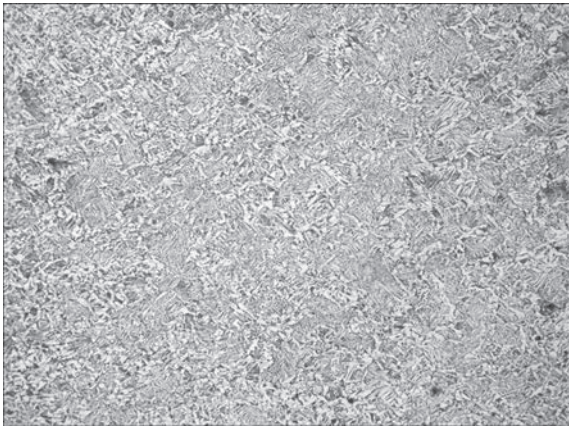


Figure 10. Weld line and overheated zone ($\times 100$)

the molten metal moves in the same direction, uniformly covering the entire surface of the pipe ends. During upsetting, the rotation of the molten layer is preserved, ensuring uniform redistribution of the metal in the joint zone and contributing to the formation of a low-profile, symmetrical weld.

FEATURES OF MICROSTRUCTURE FORMATION DURING MIAB WELDING

The microstructure of the welded joint during MIAB welding is determined by the combined effects of the heating temperature regime, cooling rate, and deformation conditions during upsetting. Thanks to the relatively low temperature of the near-contact layers and the high thermal cycle rate in MIAB welding, conditions are created for the formation of a fine-grained structure with uniform phase distribution.

Metallographic studies of welded joints of steel pipes with dimensions of 121 $\times$ 10 mm made of 01Star520 steel showed that, a predominantly ferrite-pearlite structure forms along the joint line. In the central part of the weld, a fine-grained structure (grain size 8–9 on the standard scale) is observed, which positively influences the combination of strength and impact toughness. The ferrite grain size corresponds to grade 9 according to the standard grain size scale.



Figure 11. HAZ, full recrystallization zone ($\times 100$)

The microhardness of the metal in the joint zone is $HV1 = 2210\text{--}2450$ MPa (Figure 10).

The overheated zone of the HAZ is characterized by a ferrite-pearlite structure. Ferrite predominantly has a lamellar morphology, while pearlite is present in a small amount within the ferrite-pearlite mixture and as isolated precipitates. The former austenite grain size corresponds to grades 5–6. The width of the overheated zone is approximately 700 μm . The microhardness in this zone is equal to $HV1 = 2280\text{--}2530$ MPa (Figure 10).

The full recrystallization zone exhibits a fine-grained ferrite-pearlite structure, with a grain size corresponding to number 9. The microhardness in this zone is $HV1 = 2210\text{--}2280$ MPa (individual values up to 2300 MPa) (Figure 11).

The base metal is characterized by a rolling texture and a ferrite-pearlite structure. The grain size corresponds to number 8. The microhardness of the base metal is $HV1 = 1990$ MPa (Figure 12).

The metal structure in the reinforcement zone is characterized by a coarse-grained structure compared to the central part of the welded joint; the former austenite grain size corresponds to grades 3–4. The structure is ferrite-pearlite, ferrite is predominantly lamellar, with a small amount of Widmanstätten ferrite. The microhardness in the reinforcement zone is approximately $HV1 \approx 2210$ MPa.

Within the reinforcement, the weld line is clearly visible, characterized by a fine-grained ferrite structure. The ferrite grain size corresponds to number 8, and the microhardness is in the range of $HV1 = 2060\text{--}2210$ MPa.

INFLUENCE OF MIAB WELDING PROCESS ON THE SERVICE PROPERTIES OF WELDED JOINTS

High impact toughness of joints produced by MIAB is critically important for pipelines operating under low temperatures and dynamic loading conditions. Exper-

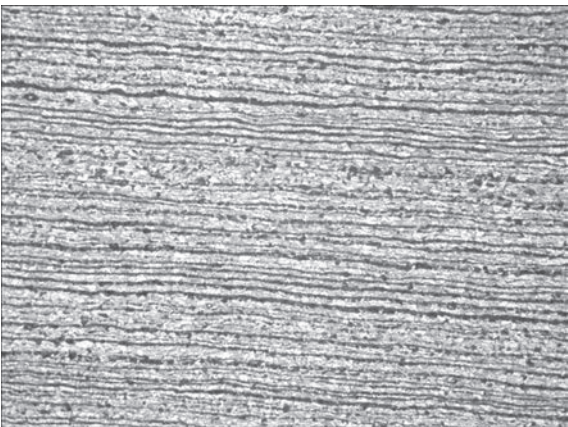


Figure 12. Base metal structure ($\times 100$)

imental results show that welded joints retain at least 80 % of the base metal impact toughness at test temperatures down to -40°C .

This combination of properties is achieved due to reduced overheating of the metal, a narrow heat-affected zone, and the absence of extended areas of structural softening. As a result, the reliability of welded joints and their resistance to brittle fracture are enhanced.

DISCUSSION OF RESULTS

The research results have shown that the stability and reproducibility of the properties of welded joints produced by MIAB welding are primarily determined by the state of the liquid molten metal in the gap between the pipe ends during the period preceding upsetting. A thickness-stable layer of molten metal ensures a uniform thermal regime, protects the contact surfaces from oxidation, and creates conditions for the complete extrusion of oxides and contaminants during upsetting.

In practice, this means that the joint quality depends not on the maximum current or arc power, but on the ability of the process to form and maintain a continuous molten layer on the pipe ends. Due to the controlled arc motion in an external magnetic field, uniform heat distribution along the perimeter and through the thickness of the edges is achieved, which reduces local overheating and prevents the formation of coarse-grained regions.

It has been found that at a certain ratio of the gap width, arc travel speed, and magnetic field induction, liquid contacts of significant area are formed in the gap. Under these conditions, the arc partially or completely extinguishes, and heating occurs in the form of conductive electrical contact heating. This combined thermal regime makes it possible to maintain a continuous molten layer even when welding pipes with wall thickness exceeding the diameter of the active arc spots, and it is a key factor in expanding the technological capabilities of MIAB welding.

The obtained temperature fields confirm that the near-contact metal layers prior to upsetting are within an optimal temperature range sufficient for plastic deformation but without overheating. This directly explains the narrow heat-affected zone, low scatter in microhardness, and high impact toughness values of the welded joints.

Thus, to ensure a stable quality in MIAB welding, it is necessary to control not only current and voltage, but primarily the parameters that determine the state of the molten metal: arc travel speed, magnetic field induction, gap width, and the timing

of upsetting. Control of these parameters enables the implementation of a reproducible welding process for thick-walled high-strength steel pipes under industrial conditions.

PROSPECTS FOR THE DEVELOPMENT OF MIAB TECHNOLOGY

Future development of welding technology is associated with improving magnetic field control systems and automating real-time process parameter monitoring. The use of adaptive control algorithms will allow maintaining stable reproduction of welded joints and expanding the range of weldable materials.

A promising direction is also MIAB process application for welding pipes and tubular components made of thermally strengthened and alloyed steels of new generations, as well as research on the possibility of welding dissimilar materials using this technology.

CONCLUSIONS

1. It has been established that a thickness-stable dynamic layer of liquid molten metal on the pipe ends during the initial stage of upsetting plays a decisive role in the formation of welded joints by MIAB welding of pipes.

2. It is shown that controlled arc motion in an external magnetic field provides uniform heating of pipe ends whose thickness exceeds the diameter of the active arc spots, and reduces the variability of the structure and properties of the welded joint.

3. Control algorithms for the parameters of MIAB welding have been developed and experimentally validated, enabling welding of pipes without the use of shielding gases or subsequent heat treatment.

4. Welded joints of high-strength steel pipes produced by MIAB welding exhibit strength and ductility characteristics comparable to those of the base metal, as well as high impact toughness at test temperatures down to -40°C .

5. A technology for MIAB welding of pipes and tubular components with diameters up to 320 mm and wall thicknesses up to 16 mm made of high-strength steels has been developed and adapted for both field and stationary operating conditions.

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CONFLICT OF INTEREST

The Authors declare no conflict of interest

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RESIDUAL WELDING STRESSES IN THE AMg6 ALLOY AFTER ELECTRODYNAMIC TREATMENT UNDER CONDITIONS OF PRELIMINARY ELASTIC TENSION

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ABSTRACT

A concept for regulating the residual stress-strain states of welded joints based on the combined use of electrodynamic treatment (EDT) of the weld metal preloaded with elastic tension (PET) has been proposed. It is shown that the use of EDT creates prerequisites for reducing the metal intensity and rigidity of the tooling for PET. It has been proved that the combined use of PET and EDT methods can contribute to the synergy of mechanisms for reducing residual welding stresses, which is based on the effect of electroplasticity realized under conditions of preliminary elastic deformation of a welded joint. It has been established that the use of EDT of the AMg6 alloy welded joints under PET conditions allows applying a 50 % lower load level σ_{PET} to reduce residual tensile welding stresses unlike without the use of EDT. It has been proven that PET of a welded joint at $\sigma_{\text{PET}} = 90$ MPa and the combined use of EDT and PET (at a similar value σ_{PET}), respectively, contribute to a reduction in residual tensile welding stresses in the AMg6 alloy by 20 and 85 %, respectively. The use of EDT creates the prerequisites for reducing the metal intensity and rigidity of the tooling for PET and will contribute to the development of methods for regulating residual stress states of welded joints based on their preload.

KEYWORDS: current pulse, welded structures, residual stresses, welded joint, electroplastic effect, preliminary elastic tension, plastic deformation, aluminium alloy, electrodynamic treatment, stress reduction

INTRODUCTION.

RELEVANCE AND AIM OF THE WORK

The need to regulate residual stresses in welded joints, which negatively affect their durability, accuracy and corrosion resistance, is one of the relevant issues in engineering practice. When manufacturing new types of welded structures, conventional methods of ensuring their optimal stress-strain states (SSS) are not always applicable.

Conventional methods of SSS regulation can be divided into thermal and mechanical methods, which are used both during and after the welding of structures [1]. In modern production conditions, the use of thermal methods of SSS regulation, taking into account energy costs, significantly increases the cost of manufacturing metal structures. This is especially true for large-sized metal-consuming [2] structures, as well as products made of light alloys with high thermal conductivity.

Force methods are based on the use of mechanical loads to the structure at different stages of its manufacture, which compensate for its residual SSS from welding. Reverse bending is an effective way to minimise welding deformations in products with relatively low bending rigidity, for example, when welding

flanges into thin-walled shell structures [3]. This requires a more complex welding technology, but in a number of cases it offers a more rational solution than correcting SSS of the structure after welding. However, the use of bending is only rational in large-scale production due to the high cost of assembly equipment.

Welding with preliminary elastic tension (PET) of parts is a method for optimizing SSS of thin-sheet structures, which was used in the production of large-sized stringer panels of light alloys [4]. The use of PET reduces residual plastic compressive deformations caused by welding. The disadvantage of this method is the significant metal intensity of the tooling for loading structural elements to be welded.

A promising approach, which is distinguished by its simplicity of application, low energy consumption and does not require metal-intensive equipment, is the improvement and development of pulse methods based on shock wave effects on the structure during (or after) its welding. Thus, the conventional method of peening welded joints has been developed with the application of an ultrasonic impact tool [5, 6]. It is distinguished by the wide range of control options despite a superficial effect on the metal being treated.

The results of previous studies have made it possible to establish the phenomenon of increased plas-

ticity and reduced resistance of metal to deformation due to the simultaneous action of active mechanical loading and pulsed electric current (PEC) with a density of $j \geq 1 \text{ kA/mm}^2$. The phenomenon was named the electroplastic effect (EPE) [7], and its practical application was implemented in the method of electrodynamic treatment (EDT) of welded joints. The use of EDT has opened up new opportunities for optimizing SSS and improving the accuracy of manufacturing welded structures [8]. The peculiarity of the EPE action differs from the known thermal effect under the action of PEC (Joule heating) in that it manifests itself only in deformed metal under the effect of mechanical loads [9]. At the same time, the synergy of the combined action of PEC and dynamic (impact) loading initiates the redistribution of residual stresses in metal structures [8, 10].

In [11], the mechanism of SSS formation by EDT under uniaxial tension of plane specimens of the AMg6 alloy of the Al–Mg system was investigated. It has been proven that EDT of specimens tensioned by a preload $\sigma_{pr} \approx \sigma_{0.2}$ of the AMg6 alloy provides greater relaxation of tensile stresses $\Delta\sigma_{pr}$ compared to variants $\sigma_{pr} < \sigma_{0.2}$ or $\sigma_{pr} > \sigma_{0.2}$. The results [12] showed that during EDT under PET conditions of the St3 steel welded joint specimens, their local plastic deformation began in the elastic region of metal loading, which was visually confirmed by the formation of Chernov–Lüders band patterns. The effect of PEC on the plasticity of titanium alloys was investigated by tensile tests of plane specimens at different degrees of relative deformation. It was shown that the action of PEC during loading contributes to an increase in the plasticity of alloys due to the synergy of EPE and Joule heating [13].

Taking into account the results [11–13], it should be noted that it is advisable to use EDT to welded joints produced under PET conditions. The advantages of combining methods (compared to their separate application) are based on the fact that, according to the PET force diagram, external forces are applied to

the parts before welding, which cause elastic tensile stresses. According to [11], this provides a greater reduction in stresses with EDT compared to the metal treatment without loading. However, the use of PET does not exclude the formation of residual plastic compressive deformations ε_w^{pl} during the cooling of the weld metal, although it does reduce their level. This reduces the ability of PET to ensure complete elimination of residual welding tensile SSS.

The use of the weld metal EDT promotes the formation of residual plastic tensile deformations ε_{EDT}^{pl} , the interaction of which with compression ε_w^{pl} results in a reduction of the latter. A positive effect of using EDT is a reduction in the PET force due to the formation of ε_{EDT}^{pl} , which allows optimizing the metal intensity of the force assembly tooling. Thus, EDT of welded joints realized under PET conditions should ensure a greater effect on the residual SSS of welded structures compared to the use of each method separately.

THE AIM

of the work is to investigate the effect of EDT under conditions of accompanying PET of welded joints made of the AMg6 alloy on their residual SSS.

PROCEDURE, MATERIALS AND EQUIPMENT FOR RESEARCH

The object of research were specimens of welded joints in the form of plates measuring $1100 \times 400 \times 2 \text{ mm}$ made of the AMg6 alloy, which were subjected to PET at different values of elastic load σ_{PET} in the specialized assembly and force tooling. The PET diagram and geometric characteristics of the welded joint are shown in Figure 1, *a*. After loading each PET specimen, as well as in further studies, the change in the value of the longitudinal (along the weld) tensile stress component σ_x on its outer surface was recorded by two methods: with the use of a mechanical strain gauge (based on 100 mm) and electronic speckle-interferometry using a laser interferometer (Figure 1, *b*) [14]. The error in recording stress values did not ex-

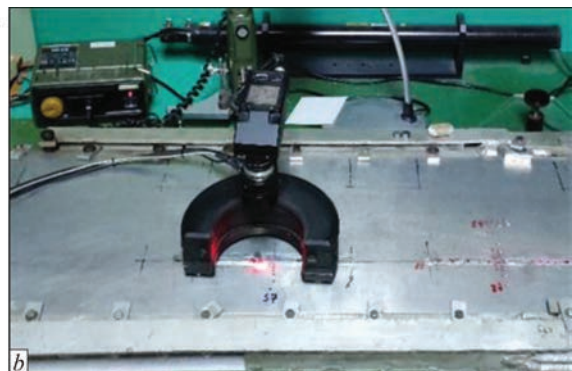
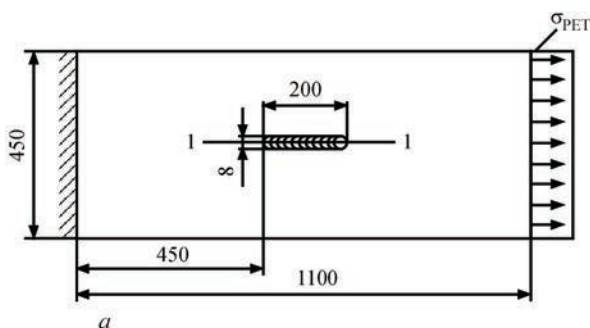


Figure 1. Preliminary elastic tension (PET) technology for the AMg6 alloy welded joint specimens: *a* — schematic diagram for PET of plates, where 1–1 denotes the line along which SSS of the plates was recorded at the stages of welding, loading, EDT and unloading, σ_{PET} is the elastic tensile load; *b* — plate with a weld and an installed interferometer, fixed in the tooling in the PET state

Table 1. Changes in SSS of the AMg6 alloy specimens $\delta = 2$ mm after PET and EDT

No.	σ_{PET} , MPa	EDT, Yes “+”, No “-”	σ_{TIG} , MPa	σ_{EDT} , MPa	σ_{unload} , MPa	$\Delta\sigma_{\text{EDT}}^*$, MPa/ %	$\Delta\sigma_{\text{PET}}^*$, MPa/ %	$\Delta\sigma_{\Sigma}^*$, MPa/ %
1	45	+	110	90	75	-20*/18	-15/16	-35/32
2	90	+	155	75	25	-80/52	-50/65	-130/ 85
3	90	-	150	-	120	-	-30/20	-30/20

Notes. 1. $\Delta\sigma_{\text{EDT}} = \sigma_{\text{EDT}} - \sigma_{\text{TIG}}$, $\Delta\sigma_{\text{PET}} = \sigma_{\text{unload}} - \sigma_{\text{EDT}}$, $\Delta\sigma_{\Sigma} = \Delta\sigma_{\text{EDT}} + \Delta\sigma_{\text{PET}}$. 2. The “-” sign indicates a decrease in stresses relative to previous values.

ceed 10 %. Given the small thickness of the test plates and excluding their vertical displacement (due to the conditions of fixing the specimens in the tooling), it was assumed that the stresses σ_x on the reverse and outer surfaces of the plates were equal. A bead measuring $200 \times 8 \times 2$ mm (with full penetration) was deposited on the surface of each elastically tensioned specimen prepared for welding to the value σ_{PET} by the TIG method using the filler wire. The deposited bead was set with geometric characteristics close to those of welded joints made of Al-based alloys.

TIG welding was performed using TIR-315 power source in the mode with the following voltage, current and process speed values: $U_w = 18$ V, $I_w = 110$ A, $V_w = 1.4$ mm/s, filler material was SvAMg6 wire of 1.6 mm diameter. After cooling and control of stress values σ_{TIG} after welding, EDT was performed on the specimen fixed in the tooling. The treatment was carried out by the manually held electrode device using Iskra-1 power source in the direction of the line along the weld centre and both fusion lines. The EDT mode was the following: $U_I = 370$ V, $U_P = 250$ V, $I_I = 4240$ A, $I_P = 1170$ A, $V_{\text{EDT}} = 4$ mm/s, where U_P , I_I and U_P , I_P are the values of voltage and current of the electric pulse (I) and shock (P) components of the electrodynamic action, respectively, and V_{EDT} is the treatment speed. After EDT, the values of stresses σ_{EDT} in the tensile specimen (Figure 1, *a*) and stresses σ_{unload} after its unloading were recorded. PET of plates was also made with the subsequent bead surfacing without the use of EDT.

The values of PET of plates and characteristics of the longitudinal SSS component at different stages of loading and EDT are presented in Table 1.

At the first stage of research, PET was performed on the specimen to a value $\sigma_{\text{PET}} = 45$ MPa ($0.3\sigma_{0.2}$ of the AMg6 alloy), followed by bead surfacing and its EDT (line 1 of Table 1). At the second and third stages, surfacing was performed after PET of the specimens to a value $\sigma_{\text{PET}} = 90$ MPa ($0.6\sigma_{0.2}$ of the AMg6 alloy), respectively, followed by EDT (line 2 of Table 1) and without treatment (line 3 of Table 1).

The results of previous studies have shown that the peak value of the σ_x component of residual tensile welding stresses for the AMg6 alloy is 120 MPa.

At the same time, the maximum efficiency of PET is achieved at values $\sigma_{\text{PET}} \approx 0.9\sigma_{0.2}$ for the AMg6 alloy [1, 2, 4]. However, in the work, PET of the specimens at values $\sigma_{\text{PET}} > 0.6\sigma_{0.2}$ of the AMg6 alloy were not used, since this would require a significant increase in rigidity and, accordingly, in the metal intensity of the tooling for tension. This contradicts the aim of the work, which consists in proving the possibility of reducing the metal intensity of the tooling for PET through the use of EDT.

DISCUSSION OF THE EXPERIMENTAL RESULTS

The distribution of the longitudinal stress component σ_x along the weld centre line during the loading–unloading and EDT stages of the test specimens is shown in Figures 2–4. From the data in Table 1, when comparing columns 2 and 4 for all variants of loading the specimens (lines 1–3), it can be seen that, in general, bead surfacing contributes to an increase in tensile stresses in the weld by 60–65 MPa due to the shrinkage in the weld metal during cooling.

EDT of the specimen in the tensile state at $\sigma_{\text{PET}} = 45$ MPa (line 1 of Table 1) contributes to a 18 % reduction in stresses— from $\sigma_{\text{TIG}} = 110$ MPa to $\sigma_{\text{EDT}} = 90$ MPa (Figure 2). This confirms the previously obtained results [11] regarding the regulation of stresses in the AMg6 alloy specimens during EDT. Further complete unloading of the specimen due to the removal of PET reduces the stress by another 16 % — from $\sigma_{\text{EDT}} = 90$ MPa to $\sigma_{\text{unload}} = 75$ MPa. Thus, the efficiency of using PET at $\sigma_{\text{PET}} = 0.3\sigma_{0.2}$ of the AMg6 alloy is close to the effect of EDT on the relaxation of residual tensile welding stresses σ_{TIG} . The combined effect of PET and EDT reduces σ_{TIG} by 32 %, i.e. $\Delta\sigma_{\Sigma} = 35$ MPa (line 1 of Table 1).

The stress state of the specimen at an increase in σ_{PET} to 90 MPa (line 2) is shown in Figure 3. The EDT of the specimen in the tensile state contributes to a 52 % reduction in stress — from $\sigma_{\text{TIG}} = 155$ MPa to $\sigma_{\text{EDT}} = 75$ MPa, which confirms the results [11]. Further complete unloading of the specimen (removal of PET) reduces the stress by another 65 % — from $\sigma_{\text{EDT}} = 75$ MPa to $\sigma_{\text{unload}} = 25$ MPa. It should be noted

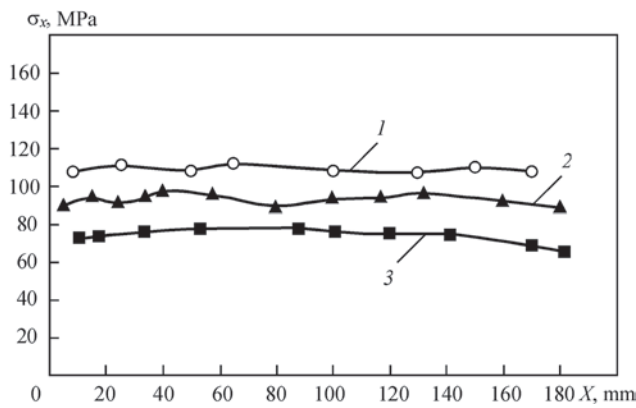


Figure 2. Distribution of the longitudinal stress component σ_x along the weld centre line during the loading–unloading and EDT stages of the specimen under the action of PET with $\sigma_{PET} = 45$ MPa (line 1 of Table 1), where: curve 1 — σ_{TIG} after PET and TIG surfacing; 2 — σ_{EDT} after EDT; 3 — σ_{unload} after unloading

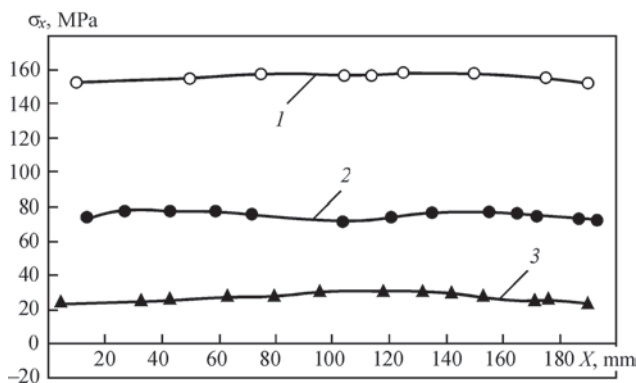


Figure 3. Distribution of the longitudinal stress component σ_x along the weld centre line during the loading–unloading and EDT stages of the specimen under the action of PET with $\sigma_{PET} = 90$ MPa (line 2 of Table 1), where: curve 1 — σ_{TIG} after PET and TIG surfacing; 2 — σ_{EDT} after EDT; 3 — σ_{unload} after unloading

that values of the relative (in %) efficiency of EDT and PET (columns 7 and 8 in line 2 of Table 1) are close and amount to 52 and 65 %, respectively. However, their absolute values differ and amount to 80 MPa (EDT) and 50 MPa (PET). Thus, EDT has a greater effect on reducing welding stresses than PET. The combined effect of PET and EDT — $\Delta\sigma_\Sigma$ at $\sigma_{PET} = 90$ MPa reduces σ_{TIG} by 85 % — by 130 MPa (line 2 of Table 1).

The stress state of the specimen at PET to $\sigma_{PET} = 90$ MPa without the use of EDT (line 3) is shown in Figure 4. Complete unloading of the plate (removal of PET) reduces the stress σ_{TIG} by 20 % — from 150 MPa to $\sigma_{unload} = 120$ MPa. In this case (EDT excluded), the value $\Delta\sigma_{PET} = \Delta\sigma_\Sigma$ and it is significantly lower compared to that obtained under similar conditions for PET of the specimen using EDT (line 2 of Table 1). Thus, the use of EDT and its exclusion at $\sigma_{PET} = 90$ MPa at close values σ_{TIG} , given in lines 2 and 3 (column 4 of Table 1), provides a reduction in residual welding stresses σ_{TIG} by 85 and 20 % from the initial level, respectively. The obtained results can be

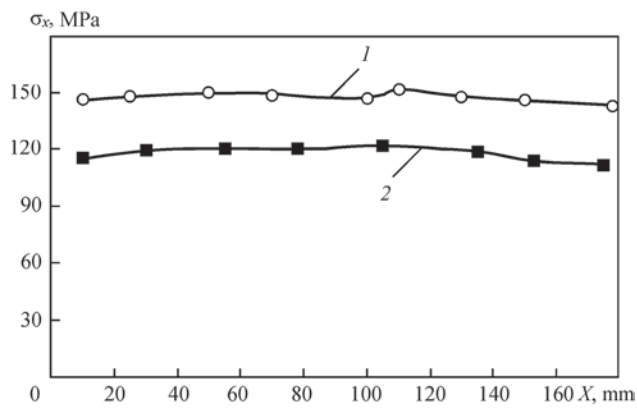


Figure 4. Distribution of the longitudinal stress component σ_x along the weld centre line during the loading–unloading stages of the specimen under the action of PET without the use of EDT with $\sigma_{PET} = 90$ MPa (line 3 of Table 1), where: curve 1 — σ_{TIG} after PET and TIG surfacing; 2 — σ_{unload} after unloading

explained by the fact that EDT (when the electroplasticity effect is realized) transforms a part of the elastic component of the stress-strain state of the weld metal into plastic tensile strains, resulting in an additional reduction in σ_{TIG} stresses.

As shown above with reference to the results of previous studies, the direction of $\Delta\sigma_{PET}$ growth may be the use of PET at σ_{PET} values close to $\sigma_{0.2}$ of the AMg6 alloy [4]. However, this approach requires a significant increase in the metal intensity of the tooling for PET and is beyond the scope of this work.

When comparing the data in Table 1, it can be seen that PET of the specimen with the subsequent EDT (line 1) at $\sigma_{PET} = 45$ MPa and PET at $\sigma_{PET} = 90$ MPa without EDT (line 3) provide close absolute values of efficiency $\Delta\sigma_\Sigma$, which are in the range of 30–35 MPa (column 9).

From the above results, it can be concluded that the use of EDT of welded joints under PET conditions allows the use of lower σ_{PET} values to reduce residual tensile welding stresses unlike in PET without EDT. Thus, the use of EDT creates the conditions for reducing the metal intensity and rigidity of the tooling for PET and will contribute to the development of methods for regulating residual stress states of welded joints using their preload.

CONCLUSIONS

1. Based on the results of previous studies, a concept for regulating residual stress-strain states in welded joints has been proposed, which is based on the combined use of electrodynamic treatment of the weld metal preloaded with elastic tension. It has been shown that the use of EDT creates the conditions for reducing the metal intensity and rigidity of the tooling for PET.

2. It has been proven that the combined use of PET and EDT methods can contribute to the synergy of mechanisms for reducing residual welding stresses

based on the effect of electroplasticity, which is realized under conditions of preliminary elastic deformation of a welded joint.

3. It has been established that the use of EDT of welded joints made of the AMG6 alloy under PET conditions allows for a 50 % reduction in tensile stress values σ_{PET} to reduce residual tensile welding stresses unlike without EDT.

4. It has been proven that PET of a welded joint at $\sigma_{\text{PET}} = 90$ MPa and the combined use of EDT and PET (at a similar σ_{PET} value) contribute to a reduction in residual welding stresses in the AMG6 alloy by 20 and 85 %, respectively.

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CONFLICT OF INTEREST

The Authors declare no conflict of interest

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MODERN TECHNOLOGIES FOR WELDING THICK-SECTION STEELS

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ABSTRACT

At present, welded structures made of thick-section steels are widely used in construction, mining, metallurgical, energy, and other industries. An analysis of existing steel welding technologies, including those employing high energy density sources, has been carried out. It is shown that the application of electron beam, plasma, laser welding, and hybrid processes currently requires a high level of fabrication accuracy and significant capital investment in auxiliary and welding equipment. The use of multipass shielded metal arc welding with coated electrodes and mechanized gas-shielded arc welding is characterized by low productivity, inconsistent quality of fusion between the filler metal and the base metal, and the formation of interlayer defects such as slag inclusions and pores. It has been established that, for welding vertical joints, promising technologies include mechanized arc welding with forced weld formation and electroslag welding. These processes are characterized by high productivity and process stability while requiring significantly lower capital investment compared to high energy density welding methods.

KEYWORDS: thick-section steel, electron beam welding, laser welding, plasma welding, hybrid welding processes, submerged arc welding, arc welding with forced weld formation, electroslag welding, mechanical properties, techno-economic justification

INTRODUCTION

The construction of heavy industrial facilities, plants, factories, pipelines, overpasses, as well as civil engineering projects, requires the fabrication of large-scale structures by joining assemblies and components of medium and large thicknesses (over 20 mm), both in workshop conditions and during field installation.

Since the beginning of 2023, demand for steel structures in Ukraine increased, primarily due to the implementation of projects aimed at protecting critical infrastructure, as well as the restoration of infrastructure and commercial facilities. The required procurement volumes of thick plate steel (more than 40 mm) proved to be so significant and sudden that a shortage emerged in the European market [1, 2]. This trend continued in 2024.

In the design and manufacture of welded structures the price-quality ratio is the primary evaluation criterion of the produced items, which can be comprehensively characterized as efficiency. Labor productivity and welding efficiency largely depend on the correct selection of the technological solution and optimal adjustment of the equipment used.

OBJECT AND SUBJECT OF THE STUDY

The object and subject of this research are welding methods applied for joining thick-section metals, their main parameters, conditions, and limitations of application.

Depending on the thickness of the materials being welded, the possibility of performing work in work-

shop or field conditions, the position and geometry of the joint (horizontal, vertical, or inclined plane, straight or curved joint), accessibility of the joint for the use of specific equipment and tooling, the required quality, productivity, and efficiency parameters various welding methods may be employed: manual metal arc welding (MMA); semi-automatic and automatic gas-shielded welding (MIG, MAG); electron beam welding (EBW), plasma arc welding (PAW), laser beam welding (LBW), and their combinations; submerged arc welding (SAW) or its combinations using different types of filler materials (solid or flux-cored wire, strips, consumable nozzles); electroslag welding (ESW).

PURPOSE AND OBJECTIVES OF THE STUDY

The purpose of this research is to analyze existing welding technologies for thick-section metals in various spatial positions in order to select the optimal welding method that ensures high-quality welded joints and economic efficiency in the fabrication of various metal structures.

RESEARCH METHODS

The methods of this study include two levels:

1. Empirical level— search and accumulation of scientific information using search engines and databases, bibliographic research and manual data collection on the subject under study.

2. Theoretical level — examination, analysis, and comparison of existing modern high-efficiency welding technologies for thick-section metals.

RESEARCH RESULTS

Let us consider the main methods applied for welding thick-section components, their principal parameters, conditions, and limitations of application.

ELECTRON BEAM WELDING (EBW)

One of the methods for welding thick-section metals is single-pass electron beam welding (EBW). This method is mainly used for welding refractory and chemically active materials [3]. Its application has been reported for heat-resistant steels 10Kh9MFBA and 10Kh12M with a thickness of 30 mm [4], as well as for super duplex steel grade 2507 with a thickness of 20 mm [5]. The quality of the weld and the possible occurrence of defects in EBW of thick metals are associated with the specific features of weld formation in different spatial positions and with joint design characteristics [3].

It should be noted that the high energy concentration in the beam leads to the formation of a characteristic weld shape in EBW (Figure 1), featuring minimal heat-affected zone (HAZ) dimensions, a high depth-to-width ratio (up to 20:1 or more), and insignificant thermal deformation of the product [6].

Despite significant achievements, the full potential of the EBW process has not found widespread application in welding thick sections and large structures due to limitations associated with operation under high vacuum conditions, where the entire welded structure must be placed inside a vacuum chamber [8].

In recent decades the new technological possibilities of chamberless EBW have been developed [9, 10]. Figure 2 shows a weld produced in the flat position on low-alloy steel using a pulsed non-vacuum electron beam.

It should be noted that the fusion zone exhibits nearly parallel sides with a well-rounded termination. With the exception of minor porosity, the fusion zone was satisfactory. Welding was performed at an average power of 26.3 kW and a welding speed of 480 mm/min.

Despite the advantages of the EBW process (high welding speed and low consumption of filler material), obtaining a high-quality weld requires careful edge preparation and the use of expensive equipment, as well as, in most cases, the necessity of a vacuum chamber, which limits the dimensions of the welded component. In addition, welding can only be performed in the flat position or in the horizontal position on a vertical plane, which further restricts the application of this method.

LASER BEAM WELDING (LBW)

Unlike EBW, laser beam welding does not require a vacuum and can be performed in air [11]. It also



Figure 1. Macrosection of a welded joint produced by electron beam welding in vacuum [7]

offers several distinct advantages over other welding processes:

- minimal shrinkage and distortion;
- high potential for automation at high travel speeds;
- high depth-to-width ratio (deep penetration with a narrow weld bead);
- small HAZ;
- applicability to components of irregular shape or in hard-to-reach locations.

Like all welding processes, laser welding must be carefully controlled to ensure quality. Poor process control may lead to cracking, porosity, incomplete penetration, or lack of fusion — any of which may result in weld failure. Over the past decade, modern solid-state and fiber lasers have become available, enabling single-pass welding without distortion of low-alloy and carbon structural steels with wall thicknesses up to 10 mm and more, at high welding speeds (approximately 2 m/min) [12].

In [13], the authors present data on deep penetration welding of structural steel S355 EN 10025 with



Figure 2. Weld produced in the flat position on low-alloy steel using a pulsed non-vacuum electron beam (NVEB), penetration depth 22 mm [7]

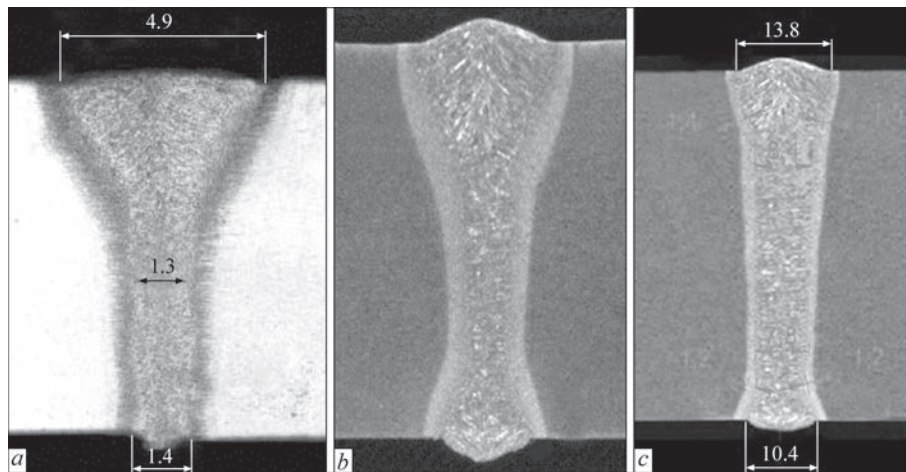


Figure 3. Penetration characteristics in laser welding of steels using CO₂ lasers: *a* — steel, thickness 10 mm, $P_{rad} = 10$ kW, $V_w = 1$ m/min; *b* — steel, thickness 30 mm, $P_{rad} \approx 28$ kW, $V_w = 1$ m/min; *c* — steel, thickness 50 mm, $P_{rad} \approx 55$ kW, $V_w = 1$ m/min [14]

thicknesses of 20 and 25 mm using a high-power fiber laser (12–30 kW).

In [14] it is shown that the penetration process of metals of various thicknesses can be successfully achieved with satisfactory weld formation and specified geometric parameters (when optimal laser welding parameters are maintained on CO₂ lasers) (Figure 3).

The authors of [13] noted that ultra-high-power laser welding (UHPLW) with laser power of 10 kW and above encounters several problems when welding thick plates (more than 10 mm), such as spatter formation, root humping, porosity, and cracking during weld cooling. In addition, heterogeneous microstructure and reduced mechanical properties remain complex and not fully resolved issues, hindering the widespread application of laser welding technologies for thick plates and requiring further research to address these challenges.

PLASMA ARC WELDING (PAW)

Plasma arc welding offers advantages over laser beam welding (LBW) and electron beam welding (EBW) in terms of equipment cost, lower joint preparation requirements and operational convenience [15].

For welding metal thicker than 6 mm, the “keyhole” mode is used [16]. In this mode, the base metal is partially melted under the influence of the high heat flux from the arc, while the surface of the weld pool is progressively pushed downward by arc pressure, leading to the formation of a blind keyhole in the base metal, followed by complete penetration that forms an open keyhole [17]. References [15, 18, 19] provide data on obtaining high-quality welded joints of steel up to 12.7 mm thick using this method, AISI 304HCu stainless steel up to 9.0 mm thick [20], as well as steels of grades SS 304 H with thickness 6 mm [21] and 316 [22].

A variation of keyhole welding is the TIG welding process developed by Fronius, known as ArcTig. The significant differences between this process and plasma keyhole welding are as follows: instead of using plasma gas, ArcTig technology employs highly efficient cooling over a wide area to concentrate the arc. This reduces the temperature at the tungsten electrode, resulting in a highly concentrated and therefore efficient arc with high energy density. The process is used for single-pass welding [23]. One of its advantages is the absence of the need for edge preparation. This process can be applied for welding materials with sheet thicknesses up to 10 mm.

Despite the fact that fundamental studies on the formation and stability of the keyhole have been presented in numerous scientific publications [24–26], the problem of ensuring stable formation of high-quality welded joints has not been completely resolved, especially for welding in positions other than the flat position.

HYBRID WELDING PROCESSES

Among advanced welding processes that allow wide control of steel penetration and ensure high productivity are hybrid welding methods, such as laser-arc welding [27, 28] and its variations: laser-arc welding under flux [29]; laser-arc welding with a consumable electrode in MIX gas [30]; and with a non-consumable electrode [31].

Despite the advantages and capabilities of hybrid welding processes, their further development faces certain difficulties associated with insufficient energy density and the complexity of process implementation at high speeds, where arc extinction may occur, resulting in unstable penetration.

More practical and economically feasible for welding thick-section steels, particularly in positions other than flat, are arc and electroslag welding processes.

MANUAL METAL ARC WELDING (MMA), SEMI-AUTOMATIC AND AUTOMATIC GAS-SHIELDED WELDING (MIG, MAG)

Despite their widespread industrial use, manual metal arc welding with coated electrodes for thick steels is mainly applied for repair purposes, when the use of other methods is impossible or less rational. This is due to low productivity and the need for additional operations (cleaning the weld metal after each pass), increased attention to quality control of the weld and HAZ, and the requirement for high welder qualification, especially when welding critical structures and performing welds in positions other than flat.

Similar issues are characteristic of gas-shielded welding methods (MIG/MAG/TIG). In addition to the disadvantages described above, protection against wind must be ensured in field conditions.

SUBMERGED ARC WELDING (SAW)

Submerged arc welding is one of the main methods for welding metals of medium and large thickness. This method allows welding of components with thicknesses ranging from 2 to 500–700 mm (single- or multipass execution) [32, 33]. Due to its high level of mechanization and automation, quality of welded joints, and high productivity, it is widely used in serial production for long welds.

However, this method is mainly applied for welding in the flat position. When welding vertical or other non-horizontal joints, additional equipment is required: forming devices, wire position correction systems, etc. Flux-cored wires are often used as filler material in submerged arc welding.

Mechanized and automated welding using self-shielded flux-cored wire or with additional CO₂ shielding allows welding in various spatial positions, ensures high joint quality, and provides increased productivity [34]. This method is widely used for thicknesses in the range of 10–25 mm, especially in field conditions.

The main technological disadvantages include the possibility of defects if the technology is violated, increased aerosol emissions, and limited visual control of the weld pool. Another negative factor is the high cost of wires, as they are manufactured in limited quantities for specific applications.

ARC WELDING PROCESSES IN SHIELDING GASES AND UNDER FLUX WITH FORCED WELD FORMATION

At the PWI, H.Z. Voloshkevych developed a method for welding thick joints with forced weld metal formation [35], which significantly increases the productivity of welding vertical joints [36]. Welding with

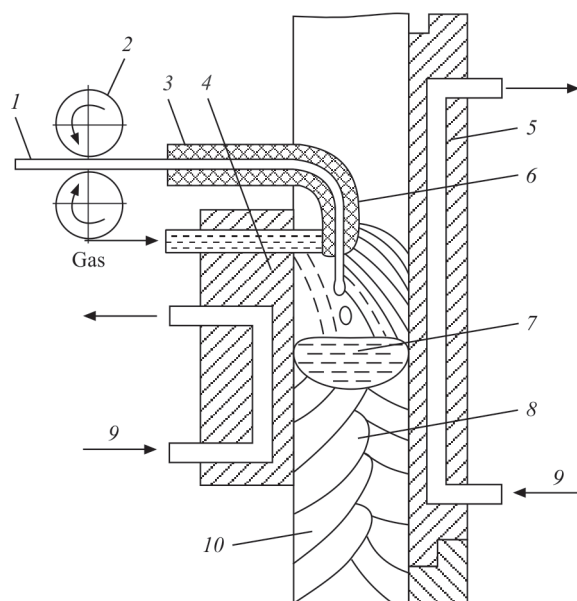


Figure 4. Schematic diagram of welding with a consumable electrode in a shielding gas of the vertical joint with forced weld formation [37]: 1 — welding wire; 2 — wire feed mechanism; 3 — current-carrying torch; 4, 5 — water-cooled copper shoes; 6 — nozzle; 7 — weld pool; 8 — solidified weld metal; 9 — water; 10 — workpiece

forced weld formation is carried out on automated welding installations and can be performed in all positions; the molten metal is retained and shaped by water-cooled copper shoes (Figure 4).

Depending on the metal thickness, welding is performed using one or several electrode wires [37, 38]. The range of weldable thicknesses is 8–40 mm. The method has been applied in the construction of tanks, ship hulls, mine supports, pipelines, I-beams, and other structures.

References [38, 39] describe the method of arc welding with forced weld formation using flux-cored wire, either self-shielded or with additional gas shielding. This process not only ensures high wire melting productivity but has also become an indispensable method for performing welds in one or two passes in various spatial positions, including circumferential welds on pipelines, gas holders, and spherical tanks, as well as overhead welding of horizontal structural belts made of metal with a thickness of at least 25 mm. The method has been successfully applied in shipbuilding. Ship hulls made of steel grades VMSt.3sp(killed) and 09G2 are welded on slipways using PP-AN5 flux-cored wire with A-1150 welding units. This makes it possible to increase installation productivity on ship hulls, improve weld appearance, and reduce residual deformations.

Welding of metal with thicknesses of 50–70 mm is performed using two solid or flux-cored electrode wires with oscillation along the joint. Carbon dioxide is used to protect the weld pool. The method is ap-

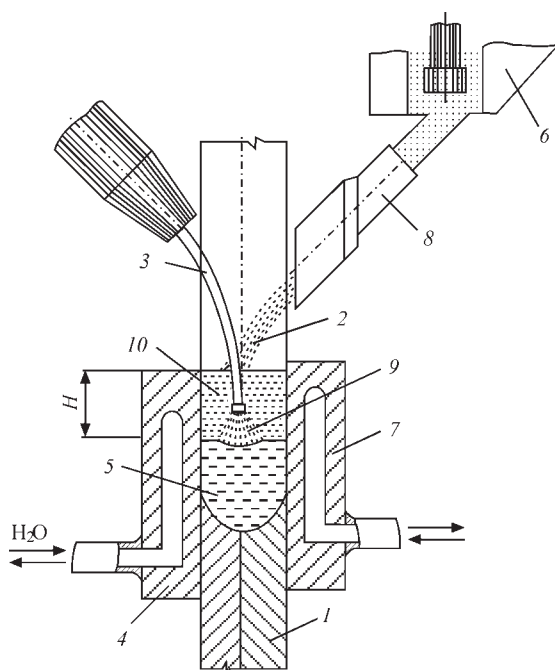


Figure 5. Schematic diagram of the welding method with forced formation of vertical welds under a thin layer of molten slag using solid cross-section wires: 1 — weld; 2 — flux; 3 — electrode wires; 4 — front sliding shoe; 5 — weld pool; 6 — flux hopper; 7 — rear forming sliding shoe; 8 — copper nozzle; 9 — welding arc; 10 — molten slag

plied in welding side hatch coamings (plate thickness 58 mm) of large container ships at the shipyard and machine-building plant of Mitsubishi Heavy Industries Limited in Kobe [40].

A method of welding with forced formation of vertical seams under a thin layer of molten slag (Figure 5) using solid-section wires was developed at the PWI [41, 42]. Studies aimed at refining the technique for performing such welds demonstrated the promise

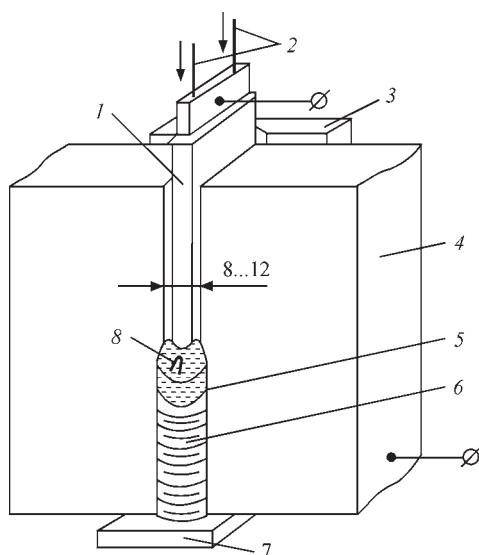


Figure 6. Schematic diagram of arc welding with a consumable insert electrode: 1 — insert electrode; 2 — filler wire; 3 — forming plates; 4 — workpiece; 5 — weld pool; 6 — weld; 7 — back-plate; 8 — arc

of this method, as it combines stable process performance with reliable protection of the weld pool.

In [43], a new method of automatic single-pass arc welding with a consumable insert electrode is proposed for specimens made of low-carbon (St3), low-alloy (09G2S, 10KhSND, 16G2AF), and high-alloy corrosion-resistant steels (08Kh18N10T, 08Kh17N13M2T) with thicknesses of 20–100 mm. The process is performed in the vertical position of the welded components and is primarily intended for welding short joints (up to 1000 mm), including under field conditions (Figure 6). The use of a plate-type electrode with an insulating coating makes it possible to reduce the gap between the welded edges to 8–12 mm, which, according to the authors, significantly improves the techno-economic performance of the metal welding process.

Despite the apparent simplicity of implementation and satisfactory properties of welded joints obtained in laboratory specimens, this method has not found widespread practical application due to the complexity of process control, which does not allow guaranteed fusion between the base and filler metals.

The described methods have found application in welding production; however, their inherent disadvantages do not allow significant expansion of their fields of application.

The methods of welding with forced formation of vertical seams under a thin layer of molten slag using solid-section wires and in shielding gas have the following drawbacks:

- absence of a compact two-wire feed mechanism suitable for field conditions that would ensure reliable and stable wire feeding into the weld pool;
- absence of a mechanism for positioning the wires within the gap between the welded edges;
- the AN-47 flux used in the process did not ensure stable welding performance;
- protection conditions of the weld pool significantly deteriorate during field welding operations, especially at height, due to wind loads;
- inability to ensure high-quality weld formation when the joint deviates laterally from the vertical by more than 30°;
- arc radiation prevents visual observation of the positioning of the welding wire between the edges.

ELECTROSLAG WELDING (ESW)

Reference [44] describes that, for welding vertical butt joints of blast furnace shells, a new electric fusion welding method was developed and successfully implemented for the first time in world practice, called electros slag welding (ESW) (Figure 7). This method was distinguished by a number of technical and tech-

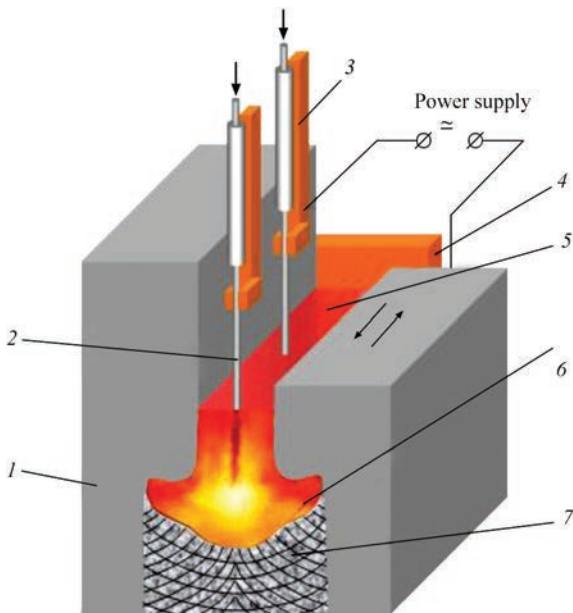


Figure 7. Schematic diagram of ESW with electrode wires: 1 — welded edge; 2 — electrode wire; 3 — current lead; 4 — water-cooled shoe; 5 — slag pool; 6 — metal pool; 7 — weld

nological advantages compared to existing industrial fusion welding methods [45].

In the manufacture of heavy machinery equipment, large welded structures in construction, shipbuilding, and other industries, two ESW methods have been most widely applied: electroslag welding with wire electrode (WEESW) and with a consumable nozzle (FNESW). These methods are used to produce the main types of joints (butt, corner, and T-joints) and weld types (straight, circumferential, variable profile) with thicknesses of 40–200 mm.

At PJSC “NKMZ,” in cooperation with the PWI, the world’s only FNESW installation was developed (Figure 8), capable of welding metal with a weld cross-section reaching 4000×6000 mm [46]. Over the past nearly 20 years, unique metal structures have been welded using this installation [47], including rolling mill stands and other heavy machinery equipment (Figure 9).

Typical areas of application of electroslag welding include: joining large castings; longitudinal welds in cylindrical pressure vessels; shells of blast furnaces and oxygen converters; stiffening ribs of ship upper decks; production of large marine container ships, whose hatch openings must be manufactured from steel plates 70–100 mm thick [48]; reinforced steel grade 20MnSiV [49]; welding of circumferential joints of presses, furnaces, engine housings, wheels for asphalt machines, joints of ship plates and cast components with thicknesses of 13–400 mm in chemical, oil, marine, and power engineering installations [50].

The ESW process has a number of characteristic advantages: high productivity (it is unrivaled in terms



Figure 8. General view of the ESW installation with a consumable nozzle for joining metal, the weld cross-section of which can reach dimensions of 4000×6000 mm

of filler metal melting rate) and process stability; deep metallurgical refining of the weld metal (absence of defects such as porosity and slag inclusions); the ability to weld excessively large thicknesses; economic efficiency; low sensitivity to edge preparation quality; and relatively low cost of welding consumables.

The main disadvantage of ESW is the high heat input, which is inherent to the nature of the process itself, the dimensions of the joint and weld pool, the use of electrode wires with diameter from 3 mm, and the requirement for high power supply capacity to maintain technological parameters. This leads to



Figure 9. Welded blank of a hydraulic press crosshead. The blank weight is 150 t; the weld cross-section (indicated by the arrow), produced by ESW with a consumable nozzle, is 2490×3860 mm

grain growth in the weld metal and HAZ, resulting in reduced mechanical properties of the joint, particularly impact toughness at low temperatures. To improve these properties, high-temperature heat treatment (HTHT) is applied; however, it may be limited or even impossible due to the large dimensions of the structures or field welding conditions.

In order to expand the application range of ESW, studies are being conducted to develop methods and technological approaches aimed at reducing heat input.

Thus, reference [51] presents and describes in detail the application of consumable-nozzle ESW in a narrow gap for producing twenty welds in the support tower of a bridge located in San Francisco/Oakland. The length of each single-pass weld was 10 m, and the joint thicknesses were 100 and 60 mm. According to [52], AASHTO (American Association of State Highway and Transportation Officials, USA) recognized the narrow-gap ESW (ESW-NG) method as suitable for welding conventional bridge steels and included it in the Bridge Welding Code (AWS D1.5:2010). Modern advances in ESW, implemented in ESW-NG, are applied at construction sites of high-rise buildings and bridges.

In [53], it was shown that by using a bifilar power supply scheme, it was possible to reduce the gap width, penetration depth, and improve the energy efficiency of the process.

The power supply scheme developed by the authors of [54] significantly increased the welding speed, making it possible to achieve satisfactory joint quality without performing high-temperature heat treatment.

References [55, 56] investigated the feasibility of narrow-gap ESW for steels with thicknesses starting from 20 mm by using small-diameter wires (1.6 and 2.0 mm), modern inverter power sources, and high-speed feeding mechanisms.

CONCLUSIONS

1. An analysis of existing welding technologies for thick-section metals in various spatial positions was conducted in order to select the optimal welding method ensuring high-quality welded joints and economic efficiency.

2. The development of the scientific and technical base has made it possible to design equipment and technologies for welding steels with thicknesses of 30–100 mm using plasma, laser, and electron beam welding methods. However, the use of high energy density sources requires substantial restructuring of the entire production system: improvement of production standards, replacement of main and auxiliary equipment, retraining of personnel, and increased precision in preparation and assembly operations. This

results in additional costs and reduced economic efficiency of production processes.

3. Arc welding methods such as MMA and MIG/MAG/TIG, characterized by low productivity, non-guaranteed fusion quality between filler and base metal, and the possibility of interlayer defects such as slag inclusions and porosity, are mainly used for repair welding when the use of other methods is impossible or impractical.

4. Mechanized arc welding and electroslog welding methods occupy leading positions in terms of application volume for welding thick-section metals, particularly in the vertical position, due to key advantages such as high productivity and process stability of ESW, as well as technological capability of welding practically unlimited metal thicknesses.

5. Based on the conducted analysis, it can be stated that further research aimed at developing and improving technologies for welding thick-section metals is promising and necessary, both for the most widely used methods such as EAW and ESW, and for processes employing high energy density sources (EBW, PAW, LBW).

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CONFLICT OF INTEREST

The Authors declare no conflict of interest

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POSSIBILITIES OF FRACTAL ANALYSIS METHODS FOR WELDED METAL MICROSTRUCTURE METALLOGRAPHIC RESEARCH (REVIEW)

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ABSTRACT

The metal of welded joints, which is characterized by increased stochasticity of structural components, requires taking into account a certain number of indicators when analyzing in the “composition–structure–properties” system. Such analysis complexity is associated with both the modern metallographic analysis methods imperfection and the difference in measurement units. The non-metallic inclusions content in the metal is determined in percent, the intergrain boundaries branching — in micrometers, the grains orientation — in degrees. It is shown that when performing metallographic research based on fractal parameterization methods of analysis of low alloyed steels weld metal structure, it is possible to obtain measurement results in numerical form, which should be the basis for creating programs for describing the interaction in the “structure–properties” system.

KEYWORDS: welding, metallurgy, metal structure, fractal analysis, microstructure, non-metallic inclusions, grain boundaries

INTRODUCTION

For the vast majority of industries, steel remains the main structural material. Modern technologies for the rolled steel production, which include complex processes of metallurgical and thermomechanical processing, are aimed at implementing the process of obtaining products with predetermined properties. When manufacturing welded structures, a carefully formed system of the metal structural grains is disturbed under the welding arc heat influence and the stress fields emergence in the weld zone. The structure of the weld metal and the heat-affected zone (HAZ) is stochastic in nature and consists of several factors, including the structural grains size, their boundaries branching, the dislocations density, the content and the size of non-metallic inclusions, their distribution in the solid solution, and others. Different dimensions are used to describe these components. The non-metallic inclusions content in the metal is determined in percent, the grains boundaries branching is in micrometers, and the grains orientation is in degrees. It is not always possible to establish a mutually unambiguous correspondence between the structure and the weld metal properties. The main reason for this is that the complex structural elements influence on metal properties remains insufficiently studied today. Modern metallographic analysis standard methods are not always able to register the features of this influence. The requirement for the metallographic analysis results validity involves the expanded database forma-

tion of experimental data and the involvement of experts with extensive experience in professional work. Usually this takes a lot of time and increases the work cost. A significant contribution to solving these problems can be made by the statistical analysis methods using metallographic analysis objects parameters, but this does not reduce the painstaking work amount in forming a database. The article authors set themselves the goal of showing that fractal analysis methods, which began to gain popularity at the end of the last century, can not only significantly simplify traditional metallographic analysis, but also increase the validity of expert conclusions.

For example, as a metallographic analysis result, the structural grains size is determined exclusively by counting both high-angle grain boundaries and low-angle grain boundaries without distinction, although it is known that low-angle grain boundaries are less effective in preventing crack growth than high-angle grain boundaries [1]. From the point of view of describing the resistance to brittle fracture, steel crystals that are separated by a low-angle grain boundary can be considered as a single unit in micro-mechanical terms [2]. In addition, it is not always possible to establish a mutually unambiguous correspondence between the results of evaluating the structural grains size according to the standard [3] and the metal properties on a point scale, which is associated with the grain shape complex geometric configuration.

This problem becomes particularly relevant when conducting metallographic studies of the weld metal

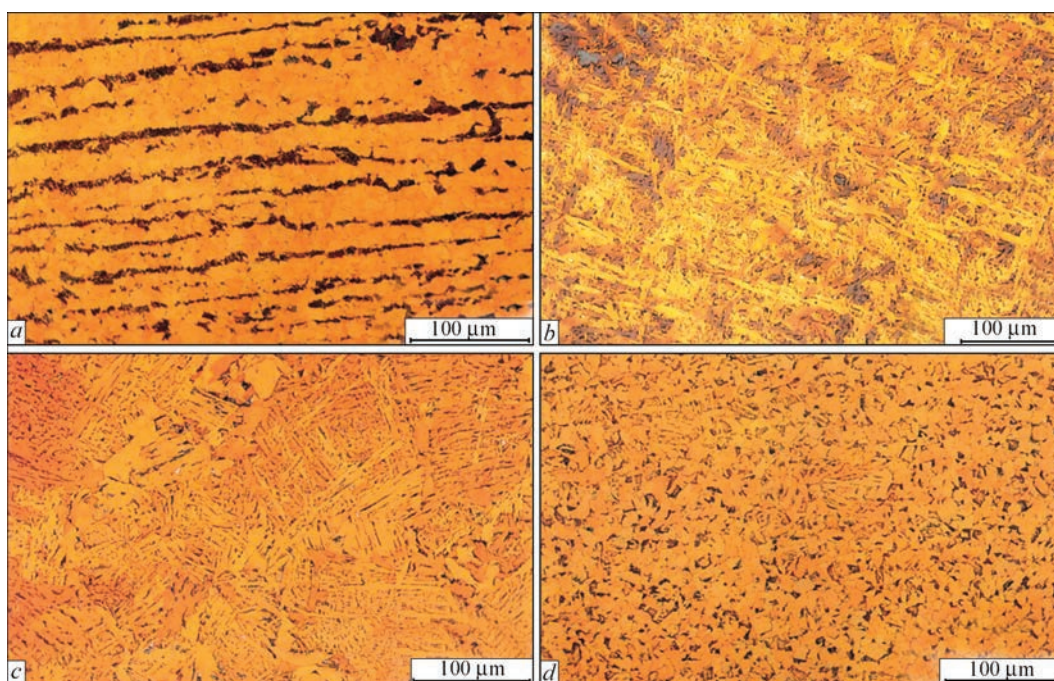


Figure 1. Microstructure of the two-pass weld metal in a butt weld of low-alloy steel sheets 25 mm thick, welded with about 3 kJ/mm input energy: *a* — base metal microstructure; *b* — weld metal microstructure; *c* — HAZ coarse-grain zone microstructure of the weld metal; *d* — HAZ fine-grain zone microstructure of the weld metal [4]

structure. From the photographs shown in Figure 1, it is clear that the weld metal and the HAZ microstructure differ significantly from the base metal microstructure. While the base metal structure (Figure 1, *a*), which is formed as a result of thermomechanical processing in rolling mills, is characterized by a certain order, both in the grains size, their location direction, and in the non-metallic inclusions precipitation places, the weld metal microstructure (Figure 1, *b*) is distinguished by stochastic formation of grains, a wide range of morphological forms of the ferrite phase with a random orientation of axis direction, the content of non-metallic inclusions of different sizes, located both in the body of the grains and at their boundaries and differs significantly from the microstructure in the coarse-grained zone (Figure 1, *c*), as well as in the fine-grained zone (Figure 1, *d*) in the welded joint HAZ.

It should be noted that traditional metallographic analysis methods do not allow us to take into account such metal structure important characteristic as its integrity, i.e. the ability to describe the metal properties through the sum of constituent elements of its structure. Such an assessment requires a statistical description of heterogeneous data large volume.

For this purpose, it is advisable to use the fractal analysis method along with other methods of diagnostics and quality control of weld metal structures, which allows unifying the indicators of structural components in numerical form. This method advantages are low cost, speed of calculations and availability. Using the fractal analysis method pos-

sibility proof for analyzing the metal structure is based on a crystallization process comparison with the geometric and stochastic fractals formation [5–7]. The published results of the metals structure fractal analysis indicate the using such feasibility method when conducting research in the “composition–structure–properties” system.

ANALYSIS OF NON-METALLIC INCLUSIONS

When performing metallographic studies in accordance with DSTU ISO 4967:1998, non-metallic inclusions in the metal are evaluated by comparison with sample images given in the standard. In this case, it is difficult to identify the specifics of the inclusions influence on the weld metal structure and mechanical properties. It is known that this influence is determined both by the inclusions size, their morphology and density of location. Inclusions up to 0.3 μm in size have little effect on the metal structure [8], inclusions compositions with a size of 0.5–0.8 μm contribute to the acicular ferrite formation with increased resistance to crack development during recrystallization [9–11], and inclusions larger than 1.5 μm in size increase the metal susceptibility to brittle fracture [12]. When analyzing inclusions, it is necessary to take into account their shape (spherical, agglomerates, films) and the nature of their location (approximately uniform distribution, accumulation, and precipitation in the chain form) [13].

In [14] it was shown that as a result of fractal analysis it is possible to not only establish the non-metal-

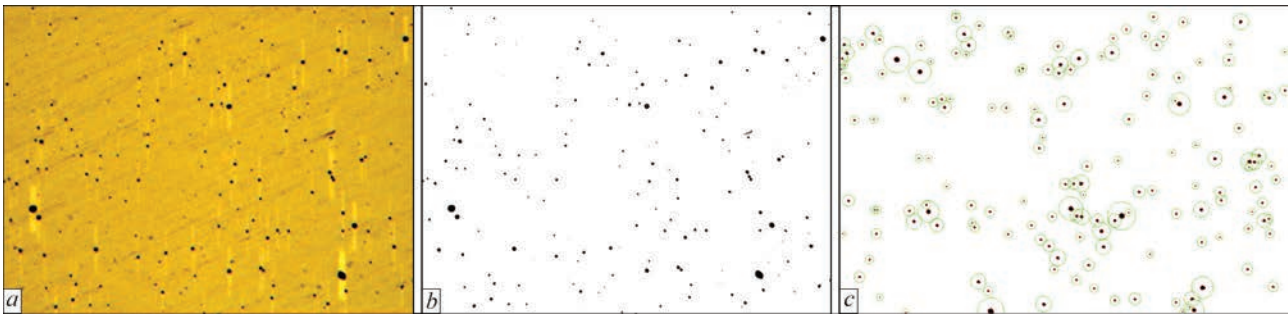


Figure 2. Sections of a low-alloy steel weld metal sample: *a* — metallographic image, $\times 500$; *b* — binarized image; *c* — identification of inclusions and their surroundings

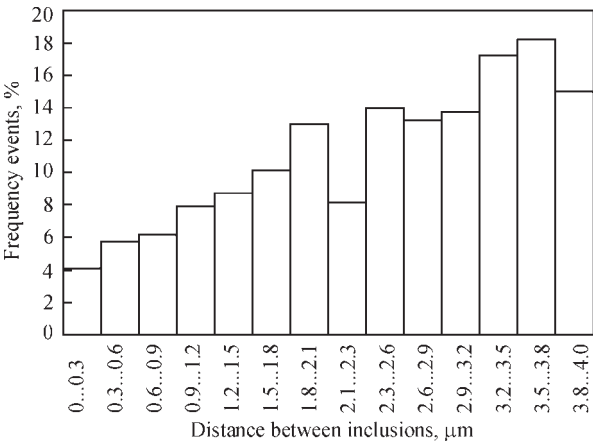


Figure 3. Distribution of inclusions over the section area, taking into account the maximum investigated distance of 5 radii from each inclusion

lic inclusions total content in the weld metal and their size distribution, but also to determine the inclusions clusters distribution density (Figures 2, 3) which can significantly increase the metal susceptibility to brittle fracture.

A significant advantage of this approach is the ability to obtain the non-metallic inclusions characteristics in numerical form, unlike traditional metallographic analysis methods, which provides grounds for increasing the analysis objectivity of metal structure various variants.

ANALYSIS OF THE GRAIN BOUNDARIES BRANCHING

When conducting metallographic studies, the structural grains size is determined in accordance with the ISO 643:2012 standard [15] by counting the boundaries through which several arbitrarily drawn straight lines pass. Such measurements results significantly depend on the metallographer qualification and require a large labor amount. Fractal analysis of weld metal structure images obtained on an optical microscope allows you to quickly determine the total boundaries length of all grains present in the image. As shown in [14], for this you need to perform a standard procedure for images binarization (Figure 4). After setting the selected gray level, the resulting image (Figure 4, *b*) allows you to measure the total grain boundaries length. This parameter is the inverse of the average grain size and, based on the results of measuring a certain images number, it can serve as an indicator for describing the weld metal structure influence on its mechanical properties.

ANALYSIS OF STRUCTURAL GRAINS

Structural grain parameters determination is one of the most common metallographic studies components. The traditional method based on the ISO 9042:1988 standard [16] is aimed at establishing the content of certain structural components in the metal. Numer-

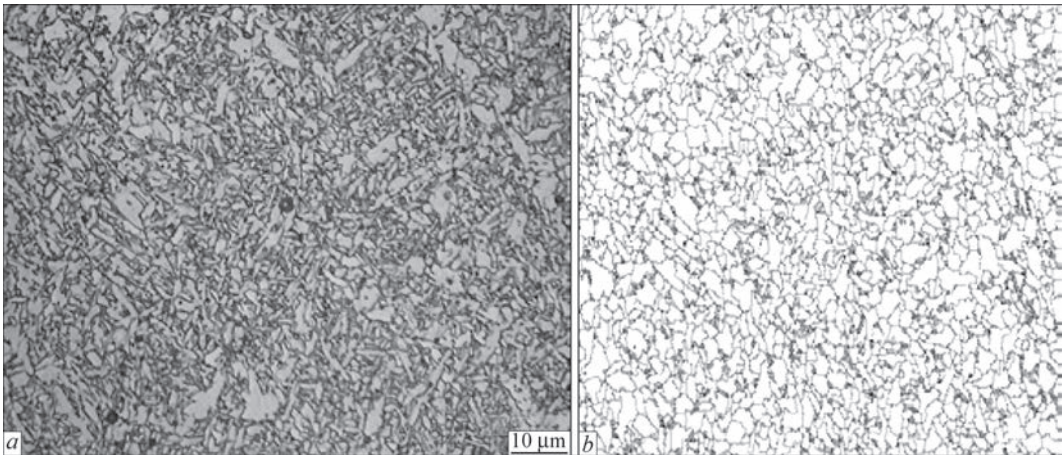


Figure 4. Sample of the weld metal microstructure: *a* — optical metallography, $\times 500$; *b* — binarized image

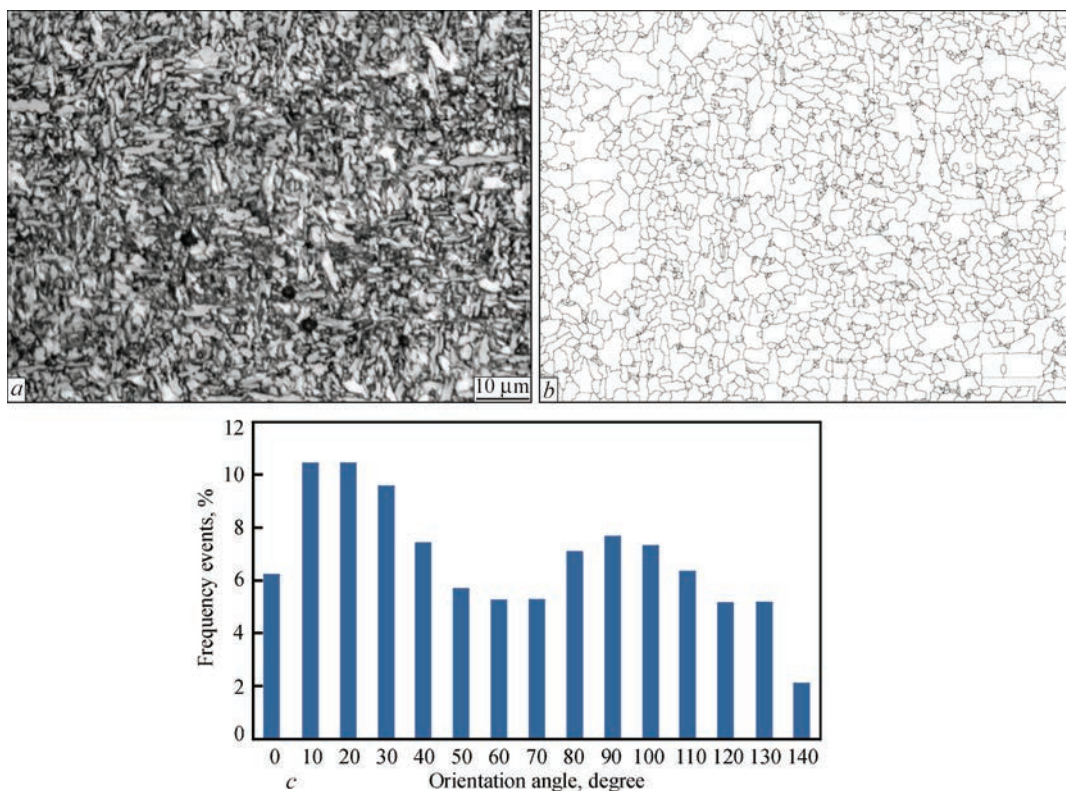


Figure 5. Results of studies of the structural grains orientation angle distribution: *a* — optical image of the weld metal microstructure; *b* — results of image processing using the MIPAR v.4.2.1 program; *c* — distribution of the structural grains orientation angle [20]

ous studies in the “composition-structure-properties” system of weld metal, which are summarized in the monograph [3], indicate the need for a more detailed structural parameters analysis. First of all, this concerns the determination of grain morphology and their orientation. Modern metallographic equipment with the installed MIPAR image analysis program (USA) uses a large library of procedures (recipes) to determine the grain size, volume fraction and size distribution of phases and inclusions, determine the structure orientation and inhomogeneity, texture features, etc.

Figure 5, *a* shows a microstructure image of a low-alloy steel weld metal obtained with a Neophot-32 optical microscope. For a clearer structural grain boundaries detection, structure optical images were first subjected to a binarization process with gray level determination (Figure 5, *b*), and then processed using the MIPAR image analysis program (USA) version 4.2.1 to determine the structural grains orientation angle. The perpendicular to the longitudinal axis of the welded joint was chosen as the given direction. The distribution of structural grains orientation angle is shown in Figure 5, *c*. Such analysis results usually allow increasing the predictability of the structure influence on the metal mechanical properties. In the rolled steel metal structure the vast majority of grains have a close orientation (Figure 1, *a*), deviations from this orientation are easy to determine from the general array and relate them to the metal mechanical

properties. The weld metal stochastic microstructure nature determines a wider grain orientation options range (Figure 5, *c*). As shown in the data presented in [16–18], the most dangerous for the brittle fracture process development are the contacting grains boundaries with a not more than 15° misorientation angle at the intergrain boundary, therefore, according to the metallographic research results, it is necessary to determine such intergrain boundaries proportion.

Research on the determination of the structure grains orientation is carried out using measuring of the fluxes of electrons reflected from the metal section surface. This method is associated with complex sample preparation and requires high metallurgical specialists’ qualifications [19]. The fractal analysis method [20], developed on the Hough transformation basis, makes it possible not only to establish the grains orientation in the metal structure photograph, which was obtained on an optical microscope during traditional section preparation, but also to determine the misorientation cases number at the intergrain boundary of contacting grains in the not more than 15° range.

Figure 6 shows an example of performing such a study [20], where the boundary between two grains with a less than 15° misorientation angle is highlighted in red, and the misorientation at the boundary of grains highlighted in green and blue is more than 30°. The results of processing a sufficient number of microstructure images using this metallographic analy-

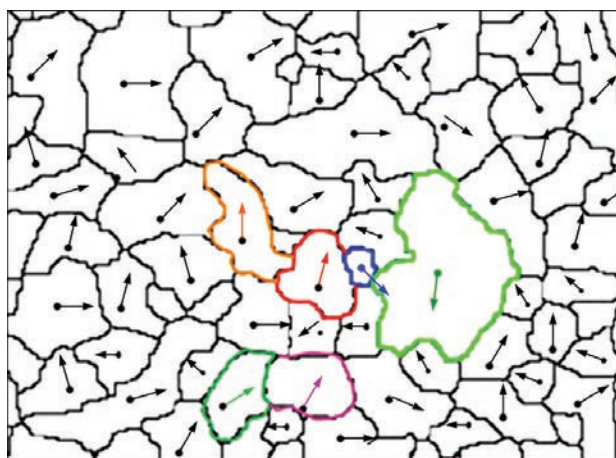


Figure 6. Examples of two neighboring grains main vector misorientation

sis method will allow establishing the boundaries proportion in the general intergrain boundaries branching that have an increased tendency to brittle fracture.

It should be noted that the data obtained as a result of research based on fractal analysis methods on various structural indicators in numerical form (non-metallic inclusions, grain boundaries branching, main grain vector orientation angle) should contribute to increasing the reliability of predicting the results of the metal structure influence on its mechanical properties.

The results of the analyses given in sections 2–4 were obtained on five weld metal samples (performed using the same technology). Three samples were selected from each weld metal for analysis. At least five optical images were obtained from each sample, each of which contained at least 10^{12} structural elements (structural grains, intergrain boundaries, non-metallic inclusions). The results of fractal analysis, which are based on such an extensive database, have a high validity level, and the analysis itself, which is performed in an automated computer processing mode, requires significantly less time than the traditional metallographic analysis method.

The possibilities of the fractal analysis method presented in the work are not enough for a thorough description of the microstructure influence on the metal properties. Such conclusions will be possible only after the accumulation of a experimental data significant base, their statistical processing and careful analysis of the obtained data. This should be the future research task, the results of which will provide an opportunity not only to expand our knowledge base on the metal microstructure formation mechanisms, but also to increase the level of HSLA steels welded joints mechanical properties.

CONCLUSIONS

Fractal analysis techniques are a progressive tool that allows increasing the metallographic analysis pro-

ductivity, end expanding the information base on the weld metal microstructure in the numerical data form.

This approach allows taking into account such an important metal structure characteristic as its integrity, that is, the ability to describe the metal properties through the sum of its structure constituent elements.

The results of structure indicators comprehensive assessment should serve as the basis for creating software for developing welding technologies for low-alloy steels with an improved level of the weld metal mechanical properties.

DATA AVAILABILITY STATEMENT

The data supporting this study's findings are available from the corresponding author upon reasonable request.

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CONFLICT OF INTEREST

The Authors declare no conflict of interest

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NEW INDUSTRIAL OPPORTUNITIES FOR ADVANCED WELDING TECHNOLOGIES

Collaboration with the scientists and engineers of the E.O. Paton Electric Welding Institute (PWI) has proven highly valuable. During a visit by Daniel Allford, President of Arc Specialties, to Kyiv several years ago, a wide range of technologies and applications was observed, many of which were either unknown or underutilized in the United States. Since that time, efforts have been directed toward identifying potential applications for processes and materials developed at the institute. Several industrial sectors have been identified where advanced technologies may provide effective solutions for challenges in manufacturing, energy, mining, and space industries.

The advent of additive manufacturing has renewed interest in high-deposition-rate welding processes, such as electroslog (ESW), which can build parts at a rate several times faster than current wire-arc additive manufacturing. Additive manufacturing of parts from exotic materials poses unique challenges. Frequently, these materials are either unavailable or impractical to produce as wire. These 3D parts are being built using plasma transferred arc welding (PTAW) with powdered metal feedstock.

The Renaissance in nuclear power has created opportunities for thick-section welding using ESW, PTAW, and electron beam welding (EBW). New technologies such as salt-cooled reactors place extreme demands on existing materials and opportunities for new solutions.

In the oil industry, as oil reserves become scarce, wells are being drilled deeper. These high-temperature, high-pressure wells test the limits of current material science and joining technologies. Welded overlays using gas tungsten arc welding (GTAW), ESW, and PTAW are used to apply metallurgically bonded coatings, which combine the strengths and properties of multiple materials to solve corrosion and strength challenges that no single material can address.

The PWI was the first, and to date, the only group to make welds in open space. Welding in space presents unique challenges, but it will be essential as humans build structures both in Earth orbit, on the moon, and eventually on other planets.

Global shortages of tungsten carbide wear-resistant coatings have created opportunities for alternative wear-resistant materials, including chromium carbide, ceramics, and man-made diamonds.

The scientists and engineers at the PWI continue to make innovations in joining, coatings, overlays, and material science, which are essential to maintain and advance the standard of living enjoyed in modern society.

FABTECH 2025: New Horizons cooperation in Welding

By invitation and with the support of the American Welding Society (AWS), as well as personally by its president Mr. Richard Holdren, and with the support of Arc Specialties represented by its president Mr. Daniel Allford, a visit to the United States was carried out by a representative of the PWI of the National Academy of Sciences of Ukraine, Dr. Volodymyr S. Kachynskyi.

The main purpose of the visit was to become acquainted with current trends in the development of welding technologies, automation, and intelligent control systems for welding processes.

During the visit to the United States, Volodymyr Kachynskyi participated in the annual AWS business meeting. The event brought together leading specialists, researchers, and executives from major companies in the welding industry to discuss strategic directions for the development of the field, the integration of digital technologies, artificial intelligence, and sustainable manufacturing.

Particular attention was paid to issues related to the training of a new generation of specialists, the expansion of international scientific cooperation, and the role of AWS as a global platform for the exchange of experience between industry, education, and science.



The annual business meeting
of the American Welding Society

hibition also highlighted social initiatives — Emerging Leaders and Women of FABTECH — focused on developing youth and female leadership in the industry.

Welding Equipment Innovations. Miller Electric presented several product premieres reflecting the industry’s shift toward mobility, digital control, and environmental responsibility:

- Millermatic 211 PRO — a MIG welder with USB upgrade capability and Auto-Set™ digital control;
- Syncrowave 212 (TIG) — a lightweight TIG unit with auto-setup and optional water cooling;
- Venture 150 S — a battery-powered welding source for field use in manual and TIG applications;
- OptX 1 kW — a portable laser welder designed to reduce distortion and spatter;
- Tregaskiss Atlas Robotic Torch — a new robotic MIG torch with improved accuracy and reliability;
- PerformArc with Intellipath™ — upgraded robotic systems with offline programming support;
- SubArc Hercules — a new high-deposition submerged-arc welding system.

These developments illustrate the company’s strategic focus on intelligent welding, higher energy efficiency, and adaptive control.

Arc Specialties (Houston, Texas) showcased its proprietary robotic welding systems and automated equipment for specialized applications — including pipe welding and complex spatial structures. The company also presented technology for electroslag surfacing of large metallic components.

Robotics and Artificial Intelligence. Yaskawa/Motoman featured a wide range of robotic welding cells — from compact ArcWorld units to hybrid systems integrating additive manufacturing.

Its new Weld Builder interface allows operators to quickly create welding programs via an intuitive control panel, reducing the time needed for robot deployment.

Many of the systems now include built-in analytics: AI-based algorithms predict defects, adjust parameters, and perform real-time diagnostics.

From September 8 to 11, 2025, the FABTECH 2025 exhibition took place at McCormick Place in Chicago — the largest industrial event in North America. This year’s show set new records: an exhibition area of 82,300 m², over 1,700 exhibitors, and more than 50,000 visitors.

FABTECH 2025 demonstrated that the welding industry is confidently moving toward the integration of digital technologies, artificial intelligence, and automation. In addition to traditional welding demonstrations, much attention was given to smart functions, data collection, and adaptive process control. The ex-





During a visit to ORNL: Dr. Adam Stevens (right), Mr. Daniel Alford (center), and Dr. Volodymyr Kachynskyi

Laser and Hybrid Welding Technologies. Laser and hybrid processes (laser + arc, laser + plasma) attracted significant attention as they combine deep penetration with precise heat control. Portable laser welders, such as the OptX 1 kW, expand the applicability of laser welding to field operations.

Ecology and Safety. A major focus of FABTECH 2025 was environmental responsibility and welder safety. Exhibitors presented modern fume-extraction and air-filtration systems, as well as next-generation personal protective equipment (PPE) — lighter, more ergonomic, and better suited for diverse operators, including women welders.

Key Development Trends. Based on the trends presented at FABTECH 2025, several strategic directions stand out:

- Hybrid processes — combining laser, arc, and plasma methods for enhanced productivity;
- Intelligent adaptation — AI-driven sensor systems for real-time parameter correction;
- Green welding — reduced emissions, improved air filtration, and safer filler materials;
- Mobility and autonomy — increasing use of portable and battery-powered welding sources.

FABTECH 2025 clearly showed that the future of welding lies in intelligent, flexible, and sustainable solutions, where robotics and digital technologies are becoming not an option, but the new standard.

Visit to Oak Ridge National Laboratory: Scientific Collaboration and Knowledge Exchange. At the invitation of Oak Ridge National Laboratory (ORNL) by Dr. Adam Stevens and support by Mr. Daniel Alford, a representative of the Paton Electric Welding Institute Dr. Volodymyr Kachynskyi visited the leading research center of the U.S. Department of Energy, located in Tennessee.

During the visit, meetings were held with ORNL's leading researchers to discuss potential areas of collaboration and to present recent scientific results from PWI. Potential areas for cooperation have been identified as electroslog welding (ESW) and electroslog additive surfacing technology, electron beam welding (EBW) and magnetically impelled arc butt welding (MIAB).

Oak Ridge National Laboratory — a Global Leader in Research and Supercomputing.

Founded during the Manhattan Project, ORNL has evolved into a world-class multidisciplinary science hub — spanning materials science, energy systems, biology, ecology, and computational research. Its most notable asset is the Oak Ridge Leadership Computing Facility (OLCF), home to Frontier (commissioned in 2022) — the world's first exascale supercomputer, capable of performing over 1 quintillion operations per second. These computing capabilities are applied to research in energy, nuclear systems, climate modeling, biomedical innovation, and advanced materials.



ORNL Research Laboratory and Research Results

Research Methods and Directions. ORNL's work combines experimental research with high-performance computing, enabling detailed study of complex physical and technological processes. Key research areas include:

- Modeling and simulation of complex systems — from nuclear materials behavior to climate dynamics;
- Development of new alloys, composites, and nanomaterials;
- Welding and repair technologies for irradiated materials used in nuclear energy.

Welding Research at ORNL. A distinctive research direction involves welding and repairing materials exposed to neutron irradiation. The main objective is to increase the durability and maintainability of structures used in nuclear power plants and thermonuclear installations.

Prospects for Cooperation. Within the framework of the visit, possibilities for joint research in the fields of welding and materials science were discussed, as well as the implementation of collaborative projects.

Today, Oak Ridge National Laboratory (ORNL) represents a unique combination of fundamental science, engineering technologies, and high-performance computing capabilities available to researchers from around the world. Innovative solutions for energy, biology, materials science, and security are developed here, while advanced methodologies in computational science enable the laboratory to maintain leading positions in the global scientific landscape.

**Daniel Allford (ARC SPECIALTIES, Houston, USA),
Volodymyr Kachynskiy (PWI, Kyiv, Ukraine)**

PATON INTERNATIONAL: INNOVATIONS SHAPING THE FUTURE OF WELDING

PATON INTERNATIONAL continues to steadily strengthen its position as one of the leading manufacturers of welding equipment, combining decades of experience with advanced technological solutions. The recent period has been particularly rich in innovations that significantly expand user capabilities and improve the efficiency of welding processes.

Welding and Cleaning in a Single Unit

One of the key innovations is the implementation of the CLEAN mode in PATON Standard-TIG and MultiPRO series machines. This solution enables the integration of two critical processes — welding and cleaning — within a single unit.

Thanks to the new mode, users can quickly clean stainless steel components and their welds from oxides and heat tint without the need for additional equipment. This reduces processing time, simplifies workflow, and enhances overall productivity.



New Product: PATON TIGCleaner



An important addition to the product line is the PATON TIGCleaner, a modern solution for the electrochemical cleaning and polishing of stainless steel products. The device has been developed with the requirements of professional manufacturing in mind, where not only joint strength but also appearance and durability are essential.

The PATON TIGCleaner ensures fast and efficient removal of oxides and heat tint formed after welding, restoring the aesthetic appearance of the surface and improving its corrosion resistance. Unlike mechanical methods, electrochemical cleaning does not damage the metal structure or affect its properties.

The unit operates in two modes: CLEANING — for removing oxides and thermal discoloration from welds, and POLISHING — for polishing stainless steel surfaces.

Key advantages include compactness and portability (weighing only 3.9 kg), ease of use, and the ability to achieve professional results without abrasive treatment.

Due to its combination of efficiency and convenience, the PATON TIGCleaner is an optimal solution for enterprises with high requirements for weld quality and product appearance.

Synergic Control: A New Level of Convenience

Another significant development is the introduction of the synergic control function for the PULSE mode in semi-automatic MIG/MAG welding across a large portion of PATON welding equipment. This solution greatly simplifies the setup of the welding process and enhances its stability.

The user only needs to set basic parameters — material type, welding wire diameter, and voltage adjustment — after which the system automatically selects the optimal operating modes. As a result, stable arc burning is ensured, metal spatter is minimized, and a uniform, high-quality weld bead is formed.

This technology is particularly effective when welding thin-sheet materials, stainless steel, and non-ferrous alloys, where precise control of heat input is critical. The combination of pulsed operation and synergic control opens up new opportunities for both professional welders and industrial enterprises.



Global Expansion

In addition to technological innovations, PATON INTERNATIONAL is actively expanding its geographical presence. The company has successfully broadened its international dealer network by entering the markets of Bangladesh and the Republic of South Africa.

To strengthen the competitive position of its products in Sub-Saharan African markets, the PATON team organized a comprehensive set of activities for representatives of importing companies and local dealers. These included a series of



training seminars and workshops, where participants had the opportunity to gain both theoretical knowledge about the extensive product range and practical skills in operating and servicing PATON welding equipment.

In addition to expanding its international dealer network, PATON INTERNATIONAL is actively developing exports through localized online channels in an increasing number of countries. Recently, local websites have been launched in the United States, Canada, India, Argentina, Brazil, Chile, Peru, Mexico, Colombia, New Zealand, and Australia.

The expansion of international presence is accompanied by the development of a partner network, enhancement of service support, and adaptation of products to meet the requirements of local markets. This approach ensures a high level of customer service and meets the expectations of professional users across different countries.

Such a strategic approach confirms the growing demand for Ukrainian welding equipment and the high level of trust in the PATON brand among international partners, strengthening the company's position as a global player in the welding industry.

More Than Equipment: The PATON Philosophy

PATON INTERNATIONAL demonstrates consistent and systematic growth by combining innovation, practicality, and a strong user-oriented approach. The expansion of equipment functionality and entry into new markets further reinforce the company's status as a modern technological leader in the field of welding.

Today, PATON represents more than just equipment — it delivers comprehensive solutions that enhance production efficiency and create new opportunities for professionals worldwide.

Anton Stepakhno (CEO, PATON INTERNATIONAL)