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THERMAL CONDITIONS OF COATED ELECTRODE WELDING OF HEAT-RESISTANT LOW-ALLOY STEELS

S.I. Moravetskyi, A.K. Tsaryuk, V.Yu. Skulskyi, M.O. Nimko

E.O. Paton Electric Welding Institute of the NASU
11 Kazymyr Malevych Str., 03150, Kyiv, Ukraine

ABSTRACT

The objective of the work consisted in determination of admissible and critical thermal modes of manual arc welding of chromium-molybdenum-vanadium steels, based on the results of metallographic investigations and measurement of weld metal hardness in as-welded and as-heat treated condition. Experimental butt welded joints of 15Kh1M1F steel were obtained under different thermal conditions. Two parameters of the welding thermal mode were set: heat input equal to 5.21; 7.78 or 10.2 kJ/cm and temperature of preheating of the metal being welded, equal to 50; 160; 270 or 360 °C. Phoenix SH Kupfer 3 KC coated electrodes of EN ISO 3580-A – E ZCrMoV1 B 4 2 H5 type were used. The subject of research was metal of the welds of the above-mentioned welded joints. For all the combinations of the heat input and preheating temperature weld metal hardness was measured after welding and after high-temperature tempering. The structure of different zones of welded joint metal was studied to determine the presence of cracklike defects and nature of the microstructure. The main attention was given to weld metal structure. The critical and admissible thermal modes of welding 15Kh1M1F steel with the selected coated electrodes were determined by the quality criteria, namely nature of the microstructure and structural homogeneity, as well as degree of weld metal defectiveness.

KEYWORDS: welded joints, welding consumables, weld metal, metallographic investigations, microstructure, manual arc welding, hardness, welding thermal mode, heat-resistant steel, cold cracks

INTRODUCTION

The range of welded assemblies from traditional chromium-molybdenum-vanadium steels of 12Kh1MF, 20KhMFL, 15Kh1M1F, 15Kh1M1FL (furtheron referred to as Cr–Mo–V steels) in steam turbines, boiler units and other components of newly designed power equipment does not show any tendency to reduction. Owing to the need for reconstruction and continuation of operation of TPP and CHPP units, which have already worked off their fleet life, the scope of repair-restoration operations with involvement of coated-electrode manual welding is also increased. Therefore, improvement of the technologies of welding steels of this type remains an important and relevant goal.

Under the conditions in the recent years, difficult for providing the industry of Ukraine with energy carriers, in the cold season the air temperature even in assembly-welding shops of machine-building enterprises can drop to below zero values. In keeping with the standards [1], Cr–Mo–V steels are allowed to be welded at ambient air temperature from 0 to –15 °C, depending on steel grade and thickness. Here, however, an accelerated drop of temperature T_p of preheating of the products being welded is in place. Insufficiently fast transition between the operations of welding, monitoring of temperature and intermediate heating combined, possibly, with the human factor, can lead to the situation, when a certain volume of the deposited metal was produced, when the actual product temperature was lower than T_p range specified by the technological documentation. Welded joints where the weld contained metal deposited with insufficient product preheating, were subjected to the required postweld heat treatment and usually successfully passed all the kinds of nondestructive testing. At the same time, weld metal hardness in a single case showed that individual measurement results exceeded the maximum admissible values, although the average hardness value satisfied the current standards [1]. Such an inconsistency is not the basis for rejection of the product, but it requires coordination with the chief materials science organization ([1], Table 18.3, items 18.4.4, 18.4.5). In view of the above-said, the technologists are trying to not just specify the optimal T_p range during welding of Cr–Mo–V steels, but at least know a very accurate practically justified bottom line of the above range, and also know the key conditions and required measures for guaranteeing compliance with current norms of weld metal hardness.

Increase of the cooling rate and lowering of the temperature level, at which γ-α-transformation runs in the metal of the weld and HAZ of alloyed steels are known to lead to increase of the fraction of diffusionless products, stresses of 2nd and 3rd kinds, dislocation density, and hardness in as-welded state. Collectively, these phenomena are known as the “structural factor”, which has an essential influence on the delayed fracture mechanism in hardened steels welded joints. So, in this case the point is that the specified minimal T_p value should be characterized not by the category of optimality, but rather criticality in terms of the risk of development of metal defects, incompatible with the welded joint serviceability.

Alongside the preheating temperature, welding heat input Q also belongs to the parameters of welding thermal mode. Welder can influence Q mainly by varying welding current I_w (within optimal ranges for each coated electrode brand and diameter). It is possible to control Q through the arc movement speed and welding technique (transverse oscillations, "stringer" beads, etc).

For welded joints of Cr–Mo–V steels required T_p is recommended or specified in a rather broad range (from 450 to 100 °C), and not always with sufficient substantiation of this range [1–5]. In a number of cases, welded joints of 15Kh1M1FL steel are also produced without preheating [1, 6]. Proceeding from data of these and many other literature sources, it did not seem possible to determine the critically necessary T_p without additional experiments.

Weld metal hardness in finished products is controlled in as-heat treated state. Known are [7] the dispersion-hardening susceptibility of metal of Cr–Mo–V steel welded joints and the temperature-time ranges of this phenomenon. This prompted an investigation into the impact of welding thermal mode on weld metal structure and properties not only in as-welded state but also after high-temperature tempering.

Thus, the objective of the work consisted in studying the influence of the welding thermal mode on the nature of structure formation and change in the hardness of weld metal of Cr–Mo–V steel welded joints in as-welded and as-heat treated state, as well as in determination of the parameters of welding thermal mode critical for ensuring the technological strength and quality of the welded joints.

MATERIALS AND PROCEDURE

Plates of 25 mm thickness cut out of a forging of 15Kh1M1F steel were taken as the base metal for experimental welded joints. This material was selected as a typical well-studied and common pearlitic steel of Cr–Mo–V system. Spectral analysis performed in the specialized PWI laboratory revealed the following content of elements in its composition, wt.%: 0.15 C; 0.39 Si; 0.68 Mn; 1.32 Cr; 1.04 Mo; 0.35 V; 0.007 S; 0.023 P, which in general complies well with the requirements of standards and specifications for 15Kh1M1F steel [8].

A large number of electrode brands were developed for welding Cr–Mo–V steels, their current production being regulated by state or industry standards of the former USSR until recently. Among electrodes of E-09Kh1MF type (GOST 9467–75), identified by marketing search via the Internet, TML-3U, TsL-20, TsL-20B and TsL-39 brands can be regarded as widely accepted in the Ukrainian market. During previous years, the validity of normative acts of the Soviet era was gradually coming to an end. For instance [9], GOST 9467–75 was replaced by STU EN ISO 3580:2019 "Welding consumables. Coated electrodes for manual arc welding of creep-resisting steels. Classification". Contrarily, the certificates of suppliers or electrode descriptions often indicate that they were made by canceled GOSTs or specifications (for instance, TU U 25.9-31230196-004:2016, TU U 25.9-34142624-014:2017). Unfortunately, plant specifications are not readily available documents, making references to them questionable in terms of modern quality assurance systems. Under the conditions of electrode production adaptation to the new standardization system introduced in Ukraine, power equipment manufacturers began to give preference to products of European companies, which have been working under the EN ISO standard system for a long time, ensuring consistently reproducible quality of the coated electrodes.

So, two brand names of electrodes of European manufacturers were pretested: Boehler Fox DCMV and Phoenix SH Kupfer 3 KC. The electrodes belong to EN ISO 3580-A – E ZCrMoV1 B 4 2 H5 type, similar to E-09Kh1MF (GOST 9467–75). By the totality of the results of previous studies (soundly determined welding technology characteristics, chemical composition of pure deposited metal), there were practically no grounds for preferring a certain brand, considering the welding production requirements. However, Phoenix SH Kupfer 3 KC was chosen for this study, for which the typical and actual C, P, Si content in the deposited metal is somewhat higher (Table 1). This is closer to the critical conditions in terms of ensuring the technological strength of the welded joint and final hardness of the weld metal.

Welding was performed in the downhand position. Electrodes were applied after their rebaking (350 °C/2 h), and directly after their cooling to room temperature. Butt joints with a V-shaped groove (20° edge bevel angle, 3 mm root face, 3 mm gap) were assembled from 25×60×300 mm plates of 15Kh1M1F steel. Weld root zone metal was produced with 4 mm

Table 1. Chemical composition of deposited metal of electrodes of EN ISO 3580-A – E ZCrMoV1 B 4 2 H5 type

Brand	Data source	Element weight fraction, %							
		C	Si	Mn	Cr	Mo	V	S	P
Böhler Fox DCMV	[10]	0.12	0.3	0.9	1.2	1.0	0.22	≤0.025	≤0.030
	A*	0.13	0.35	0.90	1.24	1.10	0.25	0.006	0.015
Phoenix SH Kupfer 3 KC	[11]	0.13	0.4	1.0	1.4	1.05	0.25	≤0.025	≤0.030
	A	0.15	0.40	0.85	1.26	1.16	0.28	0.002	0.018

*A — spectral analysis of metal of 6th layer of the deposits made in keeping with [12].

electrodes of ISO 2560-A – E 46 3 B 5 3 H5 type. Butt preheating temperature during root pass welding was 200–300 °C. The purpose of this weld area was assembling the butts and giving them the initial rigidity, so that the weld root metal was not included into the study.

Groove filling was performed with Phoenix SH Kupfer 3 KS electrodes with maintenance of a combination of Q, T_p parameters, which was exactly what characterized the welding thermal mode (the format of thermal mode designation to be used further for concise presentation is related to it: n-T_p, where n is the welding mode 1, 2 or 3, Table 2). Change of Q level was ensured by I_w variation. From the viewpoint of avoiding defects (lacks of fusion, slagging) some passes in the low, narrow area of the groove were performed with small transverse oscillations. Above it the groove was filled by just stringer beads. I_w values were assigned and maintained by VDU-506 rectifier together with welding current regulator RDE-251U3.1, fitted with an ammeter. Values of arc voltage U_a and welding speed V_w were measured directly during welding the joints.

Heat input was calculated (Table 2), using V_w values averaged by the results of measurement in six arbitrarily selected passes in each mode, as well as value η of effective efficiency of metal heating by the arc, taken to be η = 0.775 [13].

Controlled preheating of welded joint metal was performed by gas-flame method up to temperatures of 160; 270 and 360 °C. Measurements were taken using a multimeter of MASTECH MS2101 model, fitted with chromel-alumel thermocouple. In addition, welded joints were produced, welding of which was started without preheating.

In the latter case, multipass welding leads to increase of base metal temperature due to self-heating, because of relatively small dimensions of experimental butt joints. At normal speed of bead deposition, already the 2nd pass should be performed with preheating up to ~100 °C. With higher pass number T_p grows, reaching values characteristic for thermal modes of the following levels, applied in this work (160 °C and higher). To avoid it, the welded joints without preheating were produced by significantly slowing down the bead deposition. Cooling of welded joint metal after each pass to room temperature required pauses of inacceptable duration. Therefore, each subsequent pass was performed after the metal temperature has reached the level of ~60–40 °C. For formal presentation of experimental results, this case was assigned the average temperature of concurrent heating T_p = 50 °C.

In the middle zone along the weld length, microsection blanks were cut out from the produced butt joints. The blanks were subjected to the following heat treatment: at 730 °C temperature optimal for welded joints of 15Kh1M1F steel [1], and at 690 °C temperature, selection of which is hypothetically possible, based on the data of [10]. Other parameters were the same for both the heat treatment variants: time of the furnace reaching the specified temperature — 2 h; time of soaking at this temperature — 3 h 20 min (20 min for equalizing of the temperature of blank metal and the furnace); cooling to room temperature with the furnace. All the microsection blanks (12 pcs) were tempered in one placement into the furnace and were mounted in it at a sufficient distance from one another.

Microsection preparation began from surface milling for 2 mm thickness, in order to eliminate the influence of the metal surface layer, decarbonized during high-temperature soaking. The milled surface was ground and polished by materials with successive reduction of abrasive particle size. The microsections were etched in 4 % solution of nitric acid in ethyl alcohol. As-welded blanks were also subject to the necessary operations for microsection preparation. The finished microsections were used to conduct the planed studies (hardness evaluation and studying the welded joint metal structure in as-welded state and after high-temperature tempering at 730 °C/3 h).

Hardness measurements were performed by Vickers method to GOST 2999–75. Load on the indenter was 5 kg. For each microsection, HV_5 value was obtained in 3–8 arbitrary points within the upper deposited layers. Such a locality of the pricks corresponds better to the procedure of hardness measurement during acceptance testing of the real products (weld face side). A greater number of measurements were performed on the microsection in as-welded state, and a smaller number — in as-heat treated state, when the range of scattering of HV_5 values was rather narrow.

The type of the structure of welded joint metal was studied in optical microscope NEOPHOT-32 with ×25–×1000 magnification. Control measurements of microhardness were performed with PMT-3 hardness meter at 0.1 kg load on the indenter.

Table 2. Modes of manual arc welding

Mode number	Electrode diameter, mm	Current, A	Arc voltage, V	Average value of actual welding speed, 10^{-3} m/s	Average heat input value, kJ/cm
1	3.2	90–100	22	3.11	5.21
2	4.0	150–160	23	3.55	7.78
3	5.0	230–240	24	4.30	10.2

Table 3. Average values of weld metal hardness, depending on the welding thermal mode

Welding mode	Q, kJ/cm	T_p, °C											
		50			160			270			360		
1	5.21	368	230	293	366	234	289	370	238	286	334	237	299
2	7.78	357	229	289	335	228	276	327	234	282	318	236	297
3	10.2	333	225	287	332	226	279	321	230	275	318	241	294

Tone denotes weld metal hardness in the state after:	
● welding	(white)
● heat treatment at 730 °C/3 h	(dark grey)
● heat treatment at 690 °C/h	(light grey)

RESULTS OF DUROMETRIC STUDIES

Change of weld metal hardness, depending on the thermal mode and tempering temperature is of a complex nature (Table 3). In as-welded state weld metal of 363–396 HV_5 hardness was produced under experimental conditions. Increase of the amount of heat applied to the metal during welding (both due to T_p, and due to Q) leads to lowering of weld metal hardness.

High-temperature tempering at optimal temperature of 730 °C leads to lowering of weld metal HV_5 to 221–246 units. For this case, however, the nature of hardness distribution is different in that higher average HV_5 values correspond to higher T_p for each welding mode. With Q increase, HV_5 drop is preserved for all the preheating temperatures, except for 360 °C.

Now, tempering at the temperature of 690 °C resulted in an intermediate level of weld metal hardness of 268–303 HV_5. All these values are considerably higher than those of the boundary limits according to [1].

RESULTS OF METAL STRUCTURE INVESTIGATIONS

Visual observation of weld formation during welding without preheating allowed revealing the deposited metal cracking susceptibility. Under the conditions of the weakest thermal mode 1-50, starting from the middle of groove height and up to the middle of the penultimate layer, crack initiation was consistently observed almost along then entire weld length, moving from layer to layer (Figure 1, *a*). Increase of Q noticeably reduces the cracking susceptibility, but does not completely eliminate it. So, at 2-50 thermal mode short cracks were observed in individual beads (Figure 1, *b*). In case of maximal Q (3-50 mode) no cracks were visually observed during welding. Further metallographic study of the respective microsections in the latter case did reveal cracks of ~3 mm size in the weld metal along the weld height (Figure 1, *c*). In addition to cracks, which were visually observed, the metal of welds produced in weak thermal modes, contains a large quantity of microscopic discontinuities of the type of tears, the size of which is of the order of grain or subgrain dimensions (Figure 2, *a*). Similar weld defects were also found for 1-160 thermal mode (Figure 2, *b*). No such microtears were revealed in the metal of welds produced in other welding thermal modes.

No cracklike defects were found in the fusion zone or the HAZ of any the welded joints.

In as-welded state the weld metal consists of a mixture of α-phase and carbide precipitates, typical for the bainitic-martensitic structure. Structures are characterized by a great diversity, depending on the thermal mode of welding, bead position along the weld height and kind of repeated heat treatment, to which a particular area of the metal was exposed during multipass welding. The preheating temperature and heat input determine the cooling rate of welded joint metal, depending on which a structure of different dispersity and with different ratio of the components, namely bainite and martensite, as well as with different degrees of martensite self-tempering, is formed. For two "adjacent" thermal modes the structural differences of the metal can be hardly noticeable. They, however, are clearly visible for thermal modes with an essentially different amount of heat introduced into the metal (Figure 3).

Welding in 1-50 and 2-50 modes forms the most dispersed homogeneous acicular composition of martensite, which practically is in a "structureless" state (weakly etched light-colored areas), and bainite (dark-coloured areas) (Figure 3, *a*). With increase of metal heat content due to preheating and arc thermal power the changes in the weld metal are reduced to structure coarsening: morphological type of α-phase gradually changes from fine- to coarse-acicular one, degree of carbide phase coagulation becomes higher, and a more clearer network of secondary boundaries appears (Figure 3, *b*). For the cases of modes with simultaneously high Q and T_p, areas with large grains of polygonal ferrite can be found in the weld (Figure 3, *c*).

Self-tempering of martensite, forming in the range of 300–400 °C, leads to an almost complete removal of excess carbon from oversaturated α-solid solution. The degree of tetragonality of its lattice becomes so low that it is no longer measurable by X-ray structural analysis, so that the lattice almost does not differ from the BCC lattice of ferrite. In the oversaturated α-solid

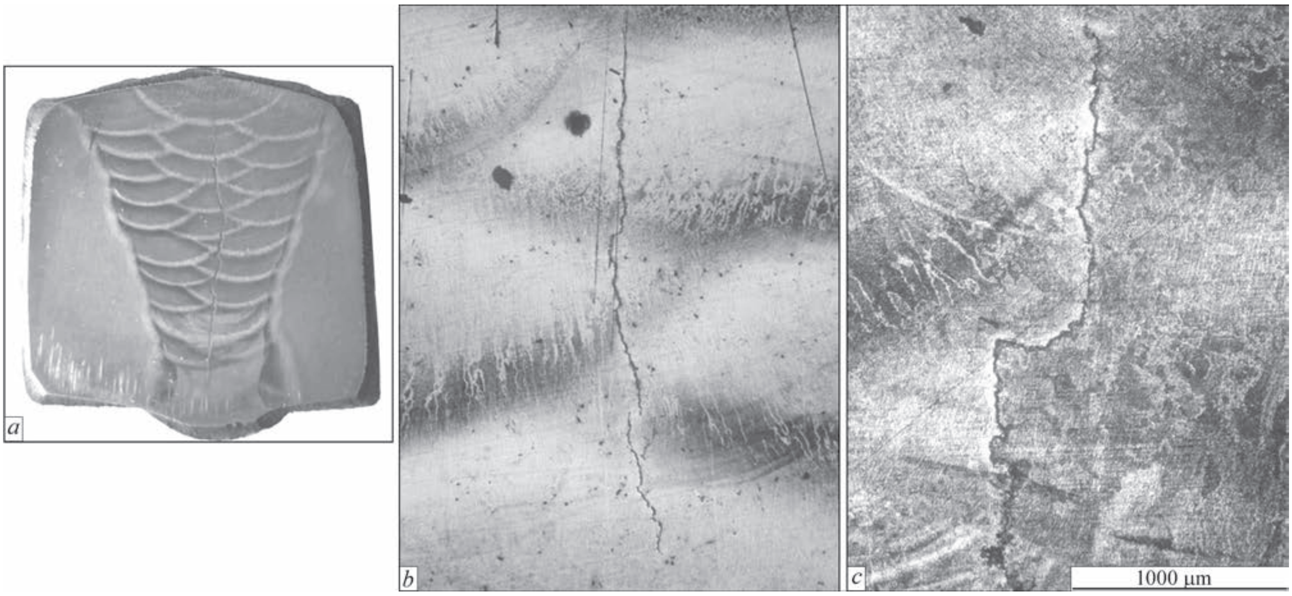

Figure 1. Cracks in the metal of welds made by welding without preheating, as-welded state: *a* — 1-50 thermal mode; *b* — 2-50 thermal mode, ×32; *c* — 3-50 thermal mode, ×50

solution, tempered at <300 °C, the degree of tetragonality of the lattice stays within the limits of sensitivity of the physical method of its evaluation [14]. Therefore, the structure of deposited metal in the state after welding with T_p = 50–270 °C contains (in addition to bainite) also martensite with different degree of self-tempering. As to the structure after welding with preheating to 360 °C, it is more appropriate to consider bainite, ferrite and carbide phases.

The structure of metal of the weld produced at $T_p >$ > 200 °C usually contains several percent of residual austenite γ_{res}, which can be present in the form of very small areas (microphases) on the grain boundaries, between bainite α-plates, and martensite laths, and thus, it can visually inconspicuous. In 200–300 °C temperature range γ_{res} decomposition proceeds with lower bainite formation [14]. Therefore, γ_{res} presence in the metal of welds made with preheating up to 270 and 360 °C is improbable.

One of the notable features of metal of all the welds observed exclusively in as-welded state, are areas of cast structure in the form of columnar crystallites, with a light-coloured fringe along their boundaries (Figure 4). The structure of the metal of light-coloured fringes along the boundaries is identified as martensitic-bainitic one, resulting from recrystallization of boundary areas in the intercritical temperature range, whereas the crystal "body" during soaking at a temperature above A_1 did not undergo recrystallization. There are the following reasons for the above identification:

- structure with light-coloured fringes around the columnar crystallites is never found in the cast zone of the last pass, or in the cast zones of the lower layers, which were not exposed to reheating up to temperatures above A_1;
- extent of such structural areas is limited (~5 mm), and their configuration follows the fusion line of the deposited beads (Figure 1, *b*);
- at a large magnification the metal structure of the light-coloured fringes (Figure 4, *b*) is not characteristic for ferrite, instead the relief shape is indicative of the product of γ→α-transformation by a shear diffusionless mechanism;
- metal microhardness is 477–518 $HV_{0.1}$ for light-coloured fringes; 384–416 $HV_{0.1}$ for middle part of columnar crystallites (data for 1-50 welding thermal mode);

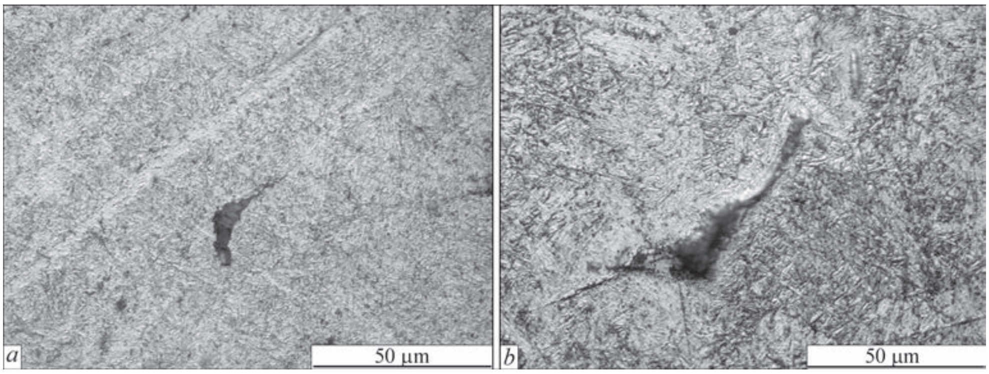

Figure 2. Microscopic discontinuities in the metal of welds made in weak thermal modes, ×1000: *a* — 2-50 mode; *b* — 3-270 mode

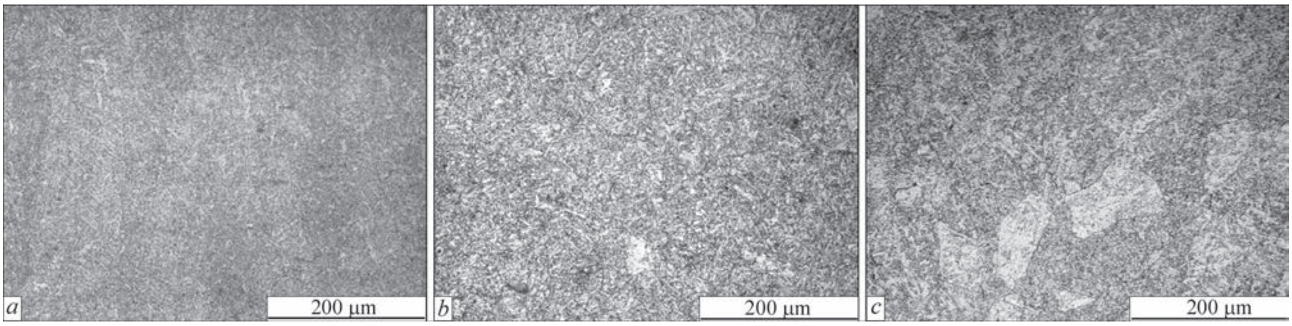

Figure 3. Type of weld metal structure, depending on thermal mode, ×250: *a* — 2-50 mode, ×100; *b* — 3-270 mode; *c* — 3-360 mode

● in as-heat treated state the light-coloured fringes are not revealed by etching, which is due to their structure type becoming similar to acicular structure of the crystallite body, due to carbide phase precipitation during high-temperature tempering.

Boundary delineation in incomplete recrystallization subzone allows evaluation of average size of the cast structure crystallites. It is interesting to note that their width depends little on the thermal mode of welding, being equal to 100 to 200 μm. Thus, increase of crystallite size with heat input rise is almost not characteristic for cast metal of this system.

Postweld tempering at 730 °C temperature leads to decomposition of oversaturated α-solid solution of martensite into ferrite and carbide precipitates. Heat treatment almost does not change the appearance of bainite areas, except for a certain increase of the quantity of the carbide phases.

A remarkable feature of the weld metal, observed exceptionally in as-tempered state, is its structural heterogeneity in the form of colonies of recrystallized ferrite grains of a large size (50–250 μm), as well as individual (single) grains of the same type in the bulk of acicular ferritic-carbide structure (Figure 5). Remnants of bainitic-martensitic structure are observed occasionally, as though locked between the adjacent ferrite grains. Proneness to this type of structural heterogeneity depends on the welding thermal mode. It is observed in the metal of welds made in thermal modes 1-50, 1-160 and 2-50. Coarse-grained ferrite is also present on macrocrack lips (Figure 5, *b*). During crack propagation, these metal areas are known to undergo intensive plastic deformation with local oversaturation by hydrogen, which tends to accumulate in the stretched zones. This leads to an assumption that ferrite appearance is associated with local deformation of metal microvolumes at the cooling stage of the welding thermal cycle, and/or during their heating at heat treatment, decomposition of unstable carbide phases in deformed volumes, hydrogen and carbon diffusion, as well as, probably, the process characteristic for hydrogen corrosion of steels (decarbonization of the solid solution due to reactions of formation of CH_4 type compounds and distribution of the latter over the crystalline lattice defects, boundaries of grains and subgrains). This kind of structural heterogeneity has a negative influence on long-term strength of the metal, because of the difference in mechanical properties of ferrite grains (163–190 $HV_{0.1}$) and surrounding dispersed acicular structure (248–268 $HV_{0.1}$) so that it is desirable to avoid it.

DISCUSSION OF INVESTIGATION RESULTS

Obtained HV_5 values of the weld metal indicate that lowering of the preheating temperature during product welding, even to 50 °C, cannot be the cause for exceeding the hardness standard values [1] (240 for the average value and 256 for individual measurements). None of the individual hardness values in the state after optimal heat treatment (730 °C/3 h), and none of the average values exceed the specified limits. Average value of 241 HV_5 for 3–360 thermal mode characterized by the highest heat input, is an exception.

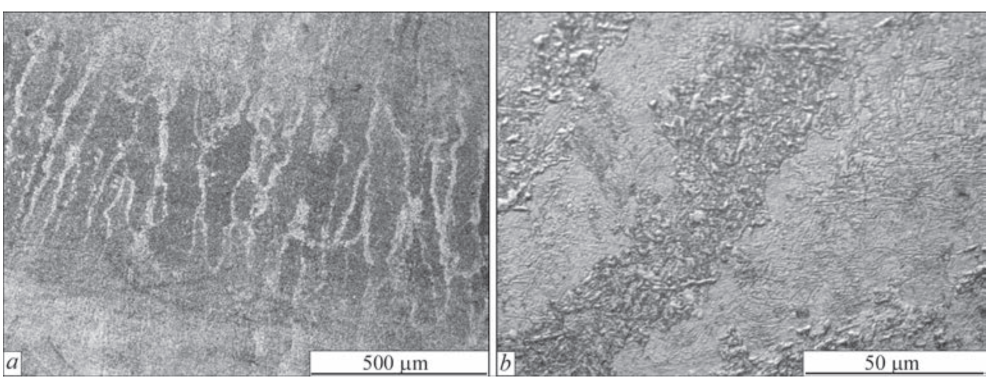

Figure 4. Structure of cast metal areas with light-coloured fringes along columnar crystallite boundaries: *a* — 2-160 mode, ×100; *b* — 3-160 mode, ×1000

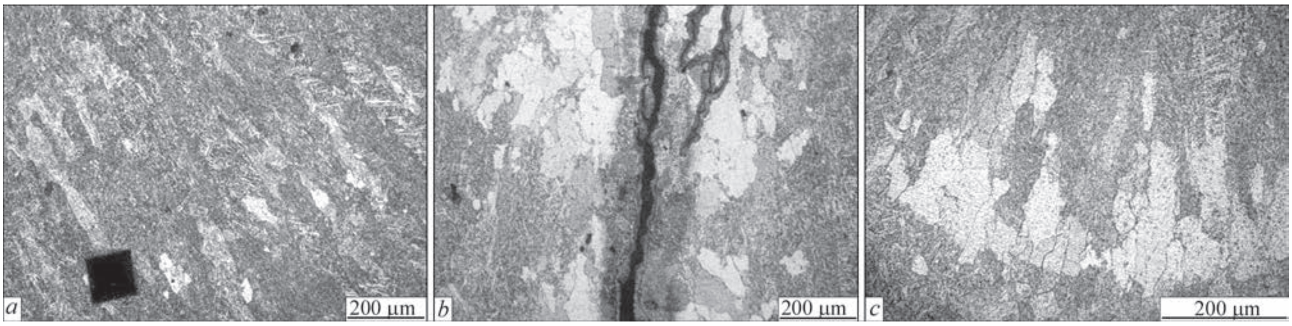

Figure 5. Structural heterogeneity of metal of as-tempered welds: *a* — individual ferrite grains in acicular structure matrix, ×125; *b* — colony of ferrite grains on the crack lips, 1-50 mode, ×125; *c* — ferrite grain colony, equidistant from the bead fusion line, 1-160 mode, ×200

An essentially higher HV_5 level for metal of all the welds tempered at 690 °C, on the whole, is an expected result for the alloying system of Cr–Mo–V steel weld, which during tempering goes through the stage of dispersion hardening. So, the noncompliance of the actual process of heat treatment of real welded products with the optimal conditions: for instance, when the product temperature (or of its part) has for a long time been in the range of 690–730 °C, not having reached the maximum of the above range, or duration of keeping at maximal temperature turned out to be shorter than the required one, should be considered the most probable cause for increased hardness of weld metal in as-heat treated state. Note that the company catalog of welding consumables [10] gives, as a reference, the level of mechanical properties and impact energy in the state after tempering by 680 °C/8 h mode for deposited metal of Boehler Fox DCMV electrodes of EN ISO 3580-A – E ZCrMoV1 B 4 2 H5 type, guaranteed by the manufacturer. However, when requirements to weld metal hardness are specified, one should not be guided by such additional information, assigning a lower temperature of welded joint tempering, or interpret them as admissibility of its random deviations in the range of 730–680 °C.

Detection of macrocracks during welding shows that level of T_p = 40–60 °C is lower than critical T_p value. Welding of Cr–Mo–V steels with electrodes of EN ISO 3580-A – E ZCrMoV1 B 4 2 H5 type is not allowed at this preheating temperature, irrespective of Q, because of unsatisfactory technological strength of the weld metal.

As is known, initiation and propagation of a brittle crack in the welded joint metal are due to the action of residual welding stresses and presence of factors, accelerating exhaustion of the metal ductility margin. One of such factors is the presence of defective areas (pores, nonmetallic inclusions, their clusters, etc.). Microtears detected in a considerable quantity in the weld made in 1-160 mode, being the ready center of fracture and mechanical stress concentration, impair the frequency distribution of the above-mentioned microscopic objects, promoting initiation and propagation of cracks. Even under the condition that these centers will not develop in the form of cracks during welding or postweld tempering, this factor even for a well-tempered metal, will probably have a marked negative influence on its long-term strength and ductility characteristics. For these reasons, 1-160 thermal mode should be also considered inadmissible.

For a more detailed determination of critical parameters of thermal mode, it is advantageous to consider the obtained results in comparison with the known quantitative characteristics, which are of criterial value as to cold cracking susceptibility of hardened steel welded joints under the welding conditions. For instance, the risk of cold cracking in the metal of a steel welded joint is considered to be high, if temperature M_e of the end of its martensite transformation is equal to 290 °C, and lower [15], and the volume fraction of martensite in its structure reaches 50 % and higher [16]. Moreover, 350 *HV* is considered to be a critical value of metal hardness, above which the risk of cold cracking becomes higher [16].

By the results of experiments (Table 3, as-welded state) average HV_5 value for the weld metal is higher than 350 units for thermal modes 1-50, 1-160, 1-270 and 2-50. For 3-50 mode, both the average and individual values are lower than 350 HV_5, which, however, was not a guarantee of crack absence (Figure 1, *c*). Similarly, 1-270 weld metal has a rather homogeneous structure without cracks or microtears, although its hardness is equal to 358–386 HV_5. This comparison shows that 350 *HV* and, probably, other proposed characteristics cannot be regarded as absolute criteria for all the possible alloying systems of hardened steels and welding conditions (welded joint rigidity, rates of heating and plastic deformation of the metal, diffusible hydrogen, etc.). We still assume that derived by generalization of a considerable volume of experimental data, these characteristics (particularly, when applied together) are suitable for tentative evaluation of cold cracking susceptibility and determination of the required conditions for their prevention during welding.

The quantity and type of final structural components in the metal of the weld and HAZ can be determined by the thermokinetic diagrams of austenite decomposition (γ), derived for continuous cooling of the metal. Let us consider (Figure 6) the diagram of 15Kh1M1FLsteel [4] in the assumption that the differences in element weight fractions do not have any essential influence on the transformation kinetics, types and quantity of their products in cast steel and in experimental weld metal.

Metal cooling rate w_{cool} during welding is one of the determinant factors of the influence on the structure of the metal of weld and HAZ of 15Kh1M1FL steel welded joints. The temperature of the start and end of transformations, as well as m values depend on w_{cool}. When w_{cool} ~ 100 °C/s, martensite is absent in the structure. At still lower w_{cool} values ($\leq$ 0.4 °C/s), precipitation of structurally free ferrite can occur along the bainite grain boundaries (undesirable manifestation of structural heterogeneity). When w_{cool} ~ 100 °C/s, the steel structure is completely martensitic. Decomposition of γ at intermediate w_{cool} values characteristic for arc welding, leads to formation of bainitic-martensitic structure. γ-α-transformation starts at temperatures of the lowest γ resistance, which, depending on w_{cool}, make up the range from 655 to 450 °C. The bainite transformation region is located higher than 400–430 °C limit (upper bainite), which, in its turn, also depends on w_{cool}. Below 400–430 °C temperatures, transformation takes place with martensite formation. The lower curve of temperatures M_e of the end of martensite transformation, which are equal from 400 to 250 °C, depending on w_{cool}, corresponds to completion of γ-α-transformations. In keeping with the diagram, critical rate of cooling during welding of 15Kh1M1F steel, when all the three criteria are fulfilled (m > 50 %; M_e = 290 °C and 350 HV), should be chosen from w_{cool} range between 25 and 36 °C/s.

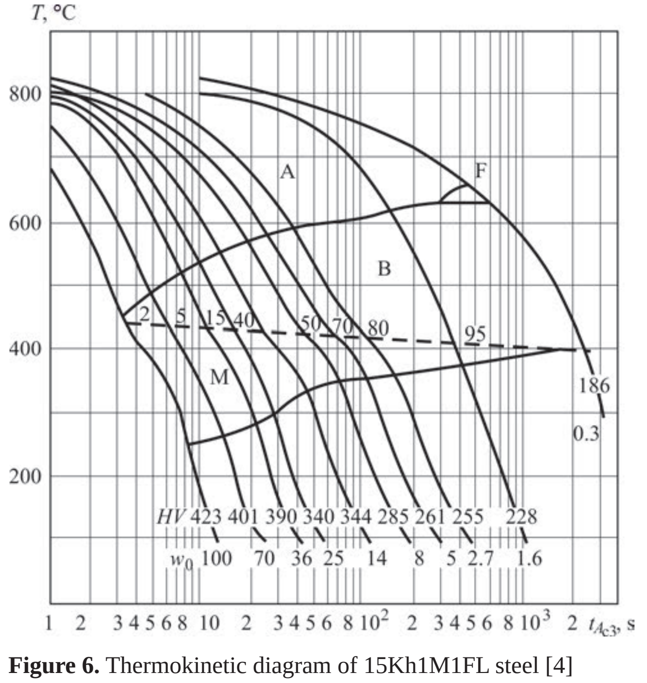

Figure 6. Thermokinetic diagram of 15Kh1M1FL steel [4]

For the conditions of manual arc welding of relatively thick-walled joints, the instantaneous cooling rate at the moment, when the metal of the weld and HAZ has reached a temperature of the lowest resistance of austenite, T_{min} can be calculated by the following formula [13]:

$$w_{cool} = 2\pi\lambda(T_{min} - T_0)^2/(q/V_w), \qquad (1)$$

where π is the pi number — 3.14159 …; λ is the metal heat conductivity, which can be taken to be 35 W/(m·°C) (for 15Kh1M1F steel in the range of 500–600 °C [8]); T_0 is the metal initial temperature (before the pass deposition), K; q is the effective thermal power of the arc, W; V_w is the arc movement speed, m/s.

Value T_{min} is determined by a curve which separates the austenite and bainite regions (Figure 6). In terms of T_{min} (w_{cool}) dependence for weak thermal modes, when high w_{cool} are anticipated (welding with concurrent self-heating to 50 °C), T_{min} was taken equal to 500 °C. For the rest of the modes T_{min} was assumed to be 550 °C. Calculation results (Table 4) allow a more accurate characterization of thermal modes. In keeping with Figure 6 and the above considerations, value w_{cool} = 33 °C/s agrees well with the concept of criticality of the respective thermal modes.

For final determination of the set of admissible thermal modes, we will take into account some other limitations, which follow from the results of this work. In terms of long-term strength of the weld metal in the state after optimal heat treatment, it is desirable to ensure a homogeneous acicular structure of the ferritic-carbide mixture. Welds produced in modes 1-270, 2-160, 2-270, 3-160, 3-270 have these characteristics. Polygonal ferrite grains in the matrix of a dispersed ferritic-carbide mixture are an undesirable kind of structural inhomogeneity. In welds made without preheating and with preheating to 360 °C, ferrite colonies and single ferrite grains are present. Taking into account this fact, as well as hardness values of 1-360, 2-360 and 3-360 welds after tempering at 730 °C/3 h (Table 3), and considering appropriate the recommendations of [3], we will limit the maximum of optimal preheating temperature by the level of 300 °C.

Current maximum admissible for 5 mm electrodes corresponds to welding at the heat input of 10.2 kJ/cm. Here, greater spattering of large electrode metal drops is observed, which firmly adhere to the plate surface and which require more time for removal. Therefore, it is not recommended to assign I_w higher than ~ 90 % of the maximal admissible one, to which Q ~ 9 kJ/cm corresponds.

Hardness of weld metal produced in 2-160 mode is lower than 350 HV, and it has a favourable dispersed structure (without macro- or microdefects or ferrite grains), however, the calculated value w_{cool} = 43 °C/s exceeds the critical one. Trying to prevent the ap-

Table 4. Calculated cooling rate of the metal at the moment of the lowest austenite resistance during multipass welding

Welding mode	Q, kJ/cm	T_p, °C			
		50	160	270	360
		$w_{cool}(T_{min}$ = 500 °C)	$w_{cool}(T_{min}$ = 550 °C)	$w_{cool}(T_{min}$ = 550 °C)	$w_{cool}(T_{min}$ = 550 °C)
1	5.21	85	64	33	15
2	7.78	57	43	22	10.2
3	10.2	44	33	17	7.8

Tone denotes w_{cool} values, corresponding to welding thermal modes:

admissible	(white)
critical	(light grey)
inadmissible	(dark grey)

pearance of undesirable features and microdefects in the production practice of welding, it is rational to somewhat increase T_p for this Q level. Required T_p (initial metal temperature) can be calculated, proceeding from (1). Taking the data of cooling graph (Figure 6): w_{cool} = 36 °C/s and T_{min} = 525 °C, we obtain T_p = 169 °C, and assume ~ 170 °C.

Nomogram (Figure 7) can be proposed as a summary of analysis of the results of this work (and taking into account the data of [3]). The area limited by *abcdef* polygon, including the dashed contour line, is the geometric locus of points, the combinations of the coordinates of which correspond to admissible values of Q and T_p during welding. The *fed* solid line is the locus of points, the combinations of the coordinates of which correspond to critical thermal modes. Below and to the left of the solid line lies the inadmissible mode region. Higher and to the right of *abcd* dashed line is the region of admissible modes, which, however, are undesirable for the above-mentioned reasons. To the left of the region of *af* contour is also the area of admissible, but undesirable thermal modes, which corresponds to low-power welding modes with a low deposition rate, poorer bead formation and higher probability of appearance of defects of the type of lacks of fusion and slagging.

Assigning one of the parameters initially, it is easy to define the optimal range and critical value of the other. Proceeding from known Q and electrode diameter, U_a, V_w parameters can be taken tentatively (Table 2), and I_w can be found from the known formula for heat input calculation [5, 13].

In the context of generalization of the work results we will formulate the main technological recommendations for producing welded joints of thick-walled assemblies from Cr–Mo–V steels with application of EN ISO 3580-A – E ZCrMoV1 B 4 2 H5 type electrodes and allowing for the probability of deviation of welded product T_p below the optimal range:

- at the stage of development of the welding technology, as well as to check the actually applied welding thermal mode for admissibility/criticality the nomogram (Figure 7) can be used together with the accompanying explanations in the text;
- 270–300 °C should be regarded as the optimal range of product preheating temperature under the ordinary conditions; at optimal preheating temperature it is possible to apply all the common methods of welding with electrodes of any diameter;
- 160 °C temperature should be regarded as the critical preheating temperature, below which performance of welding operation (bead deposition) is not allowed;
- during the cold season, when the probability of deviations of welded product T_p below the optimal range become higher, it is better to select 4 or 5 mm electrodes for welding; it is allowed to use 3.2 mm electrodes to a limited extent, or under the condition of welding at as high heat input as possible. For this purpose, it is desirable to assign the welding current near the upper admissible limit (while ensuring acceptable bead formation and spattering level), de-

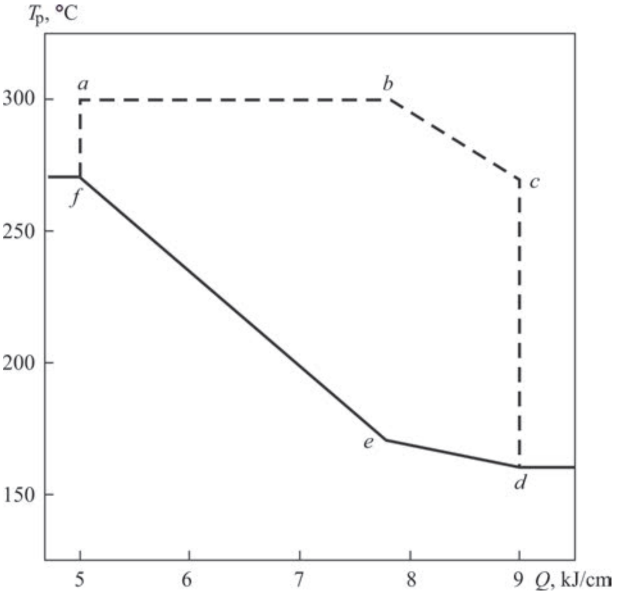

Figure 7. Nomogram of thermal modes of welding heat-resistant steels with coated electrodes of EN ISO 3580-A – E ZCrMoV1 B 4 2 H5 type

crease the arc movement speed, apply transverse oscillations of the electrode, etc. Moreover, it is allowed to perform product preheating or intermediate heating with a margin – up to temperatures somewhat higher than the optimal one (300–360 °C);

● in order to guarantee achieving an average hardness value of the weld metal after heat treatment ≤240 *HV*, excess heat input during welding (high-power modes simultaneously with metal preheating up to temperatures higher than the optimal range) should be avoided; welded joint tempering parameters should be equal to: temperature not lower than 730 °C, soaking duration of not less than 3 h; together with that it is necessary to ensure the homogeneity of heating temperature in all the product points or of the welded joint control zone (in case of local heat treatment).

CONCLUSIONS

1. The optimal range of preheating temperature of thick-walled assemblies from heat-resistant chromium-molybdenum-vanadium steels, welded by coated electrodes of EN ISO 3580-A – E ZCrMoV1 B 4 2 H5 type is 270–300 °C. Preheating temperature of 160 °C is critical under the above-mentioned conditions. Below the critical preheating temperature the probability of appearance of micro- and macrodefects of weld metal and unfavourable features of its microstructure becomes higher, which does not allow producing serviceable high-quality welded joints.

2. Meeting the hardness standards (≤240 *HV*) of metal deposited with electrodes of EN ISO 3580-A – E ZCrMoV1 B 4 2 H5 type is ensured by welding in modes without excess heat input. In addition, the following postweld tempering parameters should be ensured: not lower than 730 °C temperature, and not less than 3 h time of product soaking at this temperature.

3. Technological recommendations on producing welded joints of chromium-molybdenum-vanadium steels with electrodes of EN ISO 3580-A – E ZCrMoV1 B 4 2 H5 type, taking into account the probability of deviation of the preheating temperature below the optimal range were developed, as well as the nomogram of admissible and critical welding thermal modes, which were transferred for practical application to JSC “Ukrainian Energy Machines”.

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ORCID

S.I. Moravetskyi: 0000-0002-5807-8340,
A.K. Tsaryuk: 0000-0002-5762-5584,
V.Yu. Skulskyi: 0000-0002-4766-5355,
M.O. Nimko: 0000-0002-9672-4921

CONFLICT OF INTEREST

The Authors declare no conflict of interest

CORRESPONDING AUTHOR

S.I. Moravetskyi
E.O. Paton Electric Welding Institute of the NASU
11 Kazymyr Malevych Str., 03150, Kyiv, Ukraine.
E-mail: box.55@ukr.net

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