TECHNOLOGY OF HEAT TREATMENT OF PIPE JOINTS FROM STEEL OF K56 GRADE PRODUCED BY FLASH-BUTT WELDING

S.I. KUCHUK-YATSENKO[1], **Yu.V. SHVETS**[1], **V.F. ZAGADARCHUK**[1], **V.I. SHVETS**[1], **V.I. KHOMENKO**[2],
S.I. ZHURAVLYOV[2] and **A.Ya. SUDARKIN**[2]
[1]E.O. Paton Electric Welding Institute, NASU
11 Bozhenko Str., 03680, Kiev, Ukraine. E-mail: office@paton.kiev.ua
[2]CJSC «Pskovelektrosvar»
3 Novatorov Str., 180680, Pskov, Russia. E-mail: info@pskovelectrosvar.ru

To determine the optimum mode of heat treatment of flash-butt welded joints of pipes from K56 steel the influence of soaking time and heat treatment temperature on impact toughness and yield limit was studied. Testing was performed in keeping with API 1104 and DNV-OS-F101 international standards. Metallographic examination of the influence of microstructure on strength properties of the joints in Neophot-32 light microscope and JAMP 2000F scanning electron microscope was conducted. Analysis of the obtained data allowed establishing the optimum heat treatment mode, at which the joints meet the normative requirements made of welds at construction of critical pipelines, including off-shore pipelines. 10 Ref., 11 Figures.

Keywords: *flash-butt welding, pipe steels, pipelines, welded joints, mechanical properties, microstructure, impact toughness, heat treatment, forced cooling*

Over the recent years construction of ultrahigh-capacity pipelines for gas and oil transportation was deployed in the Russian Federation. New generations of pipelines operate at higher pressures compared to traditional working pressure (gas pipelines — 100–200 atm, oil pipelines — 75–100 atm). Here, pipe wall thickness is increased (up to 39 mm) [1]. To save metal and lower the pipeline construction cost while preserving their reliability, requirements to mechanical properties of pipe metal, in particular, ductility properties, cold resistance and weldability are increased. In order to meet the high requirements made of pipes, metallurgical companies developed steels with ferritic-bainitic and bainitic structures.

At the same time higher requirements are now made of mechanical properties of circumferential pipe joints welded in site [2]. This pertains, primarily, to impact toughness properties at low testing temperatures.

Earlier we developed [3] the technology of flash-butt welding (FBW) of pipes from steels of K56 grade of 1219 mm diameter and 27 mm thickness, used for off-shore pipelines. Pipes were manufactured in keeping with TU 14-3-1573–96. Composition of pipe metal is as follows, wt.%: 0.06 C; 0.21 Si; 1.42 Mn; 0.12 Ni; 0.07 Mo; 0.04 V; 0.04 Al; 0.02 Ti; 0.05 Cr; 0.02 Nb; 0.004 S; 0.012 P. Such a composition implies lower content of carbon, and alloying by niobium, titanium and vanadium that in combination with the methods of thermomechanical treatment [4] ensures high strength and cold resistance required in pipeline operation.

Mechanical properties of pipe metal are as follows: σ_y = 484.4–493.5 MPa; σ_t = 546.7–556.8 MPa; KCV_{20} = 334.7–336.6 J/cm^2; KCV_{-40} = 333.0–336.6 J/cm^2; hardness HV5-1850–1950 MPa.

Required set of properties is achieved due to grain refinement, dispersion, dislocation, subgranular and solid solution strengthening [5]. Steel microstructure is a ferritic matrix with a small number of bainite and carbide inclusions (Figure 1). Size of ferrite grains elongated along the rolling direction is equal to 5–10 μm, that corresponds to point 10–11.

Steels of this type are thermally unstable. Welding heating leads to microstructural changes, which result in joint mechanical properties differing from those of the base metal.

Tensile testing showed that the ultimate strength of welded joints decreases by approximately 6 %, compared to base metal, and is equal to 520 MPa. Fracture runs at the distance of about 18–20 mm from the joint plane. According to microstructure (Figure 2) the fracture site corresponds to the section of high-temperature tempering, in which the joint hardness is minimum at approximately HV5-1700 MPa (Figure 3). It

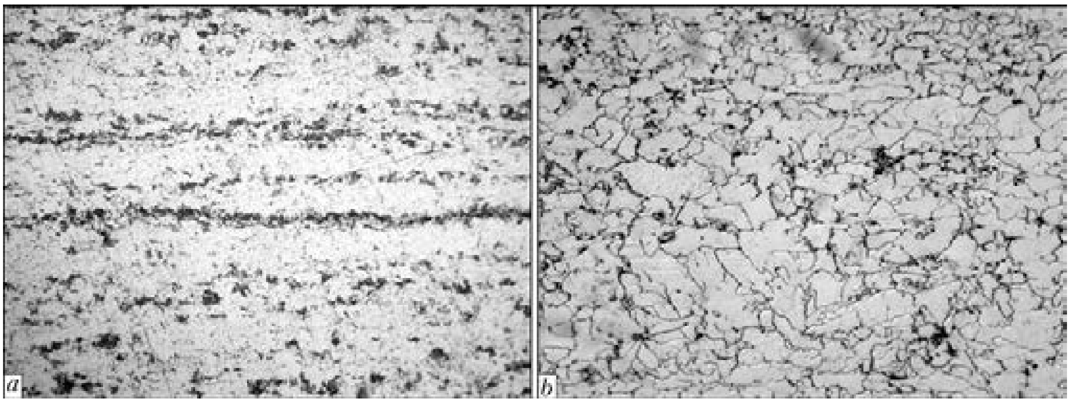


Figure 1. Pipe steel microstructures (a — ×100; b — ×400)

should be noted that directly in the butt joint zone after welding hardness is somewhat higher than that of base metal — approximately $HV5$-2100 MPa. Lowering of yield limit in the section of high-temperature tempering can be prevented by lowering the heat input in welding either by shortening the welding time, or by increasing the cooling rate.

Here bend testing showed that bend angle of all the welded joints was equal to 180 °C and the weld had no cracks. Impact bend testing turned out to be critical. Results of testing Charpy samples with a notch along the joint line show a slight lowering of impact toughness in the vicinity of the butt. While base metal impact toughness was 335.8 and 334.9 J/cm^2 at 20 and −20 °C, respectively, the average impact toughness in the section of the joint line decreased to 15.0 J/cm^2 at the temperature of 20 °C and to 8.1 J/cm^2 at −20 °C.

Achieved level of impact toughness does not meet the requirements of DNV-OS-F101 Off-Shore Standard [2].

As shown by metallographic examination, the low impact toughness is predetermined by metal microstructure in the joint zone. The structure consists of bainite with polygonal grains of proeutectoid ferrite along the boundaries of primary austenite grains (Figure 4, a), the size of which corresponds to point 3. Coarse grain and polygonal ferrite along the grain boundaries are known to be factors lowering the cold resistance [6].

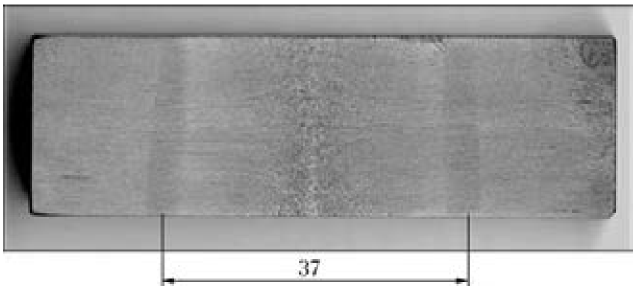


Figure 2. Macrostructure of pipe steel joints made by FBW

Increase of cooling rate after welding leads to reduction of the volume fraction of proeutectoid ferrite and greater dispersity of bainite (Figure 4, b). However, it only slightly increases the impact toughness from 6.9 up to 8.1 J/cm^2 at −40 °C.

It is known [7] that recrystallization at normalizing improves weld microstructure. It is believed to be rational to apply local high-temperature treatment in the butt area to improve the impact toughness.

Given below are the results of investigations on development of basic technology of heat treatment (HT) of FBW joints of pipes from steel 10G2FB (TU 14-3-1573–96).

Local heating of welded joints 320 mm wide with removed flash was performed with a single-turn enclosing inductor heater of 2.4 kHz frequency.

Temperature monitoring was performed by the following schematic: temperature sensor — sensor signal normalizer — analog-digital converter — computer. Multichannel modules with individual galvanic decoupling in the channels of HL-7B 30-06 type providing output signal of 0–10 V DC voltage at thermo-emf variation in the range of 0–50 mV were used as signal normalizers. Stand-alone twelve-digit analog-digital

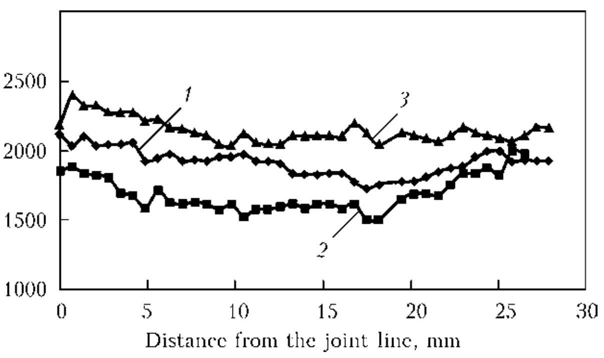


Figure 3. Distribution of $HV5$ hardness in the welded joint: *1* — after welding and cooling in air; *2* — after HT and cooling in air; *3* — after HT and cooling by water-air mixture

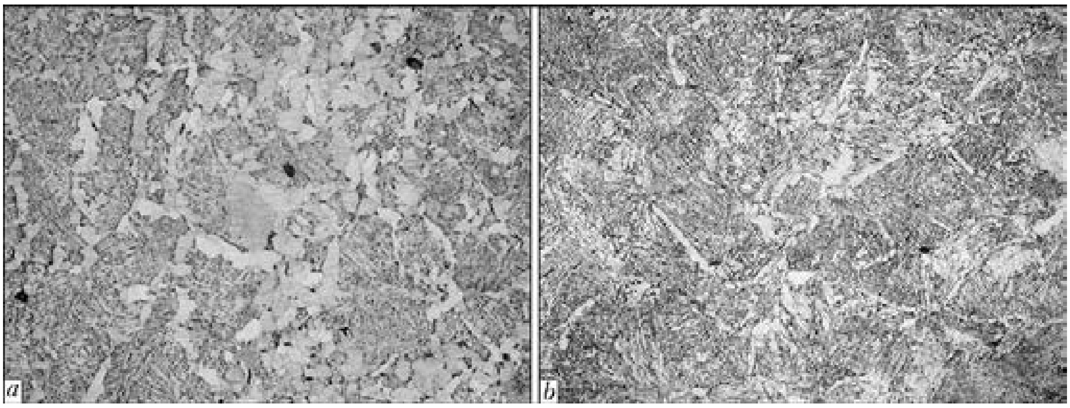


Figure 4. Microstructures (×100) of pipe steel welded joint after its cooling in air (*a*) and in water (*b*)

converter of ADS-12E type was connected to a notebook type computer vie a parallel interface. Temperature sensors were chromel-alumel thermocouples of 0.5 mm diameter, which were welded to samples by capacitor-type welding in the heating zone.

Cooling was performed either in air, or forcedly from two sides by an air-water sprayer.

Evaluation of impact toughness values of the joints after HT was conducted on Charpy samples with a sharp-notch in the joint plane, located normal to the pipe surface, at −20 °C temperature. Testing was performed in certified PWI laboratory for mechanical testing.

Metallographic examination was conducted in Neophot-32 light microscope after section etching in 4 % alcohol solution of nitric acid.

HT mode includes the following components: heating rate, heating temperature, time of soaking at constant temperature, rate of cooling after soaking.

Criterion for selection of the heating rate is producing a uniform temperature field by sample thickness. Conducted investigations revealed that uniform heating of butt joints up to 950 °C temperature across the entire thickness is achieved in 5 min.

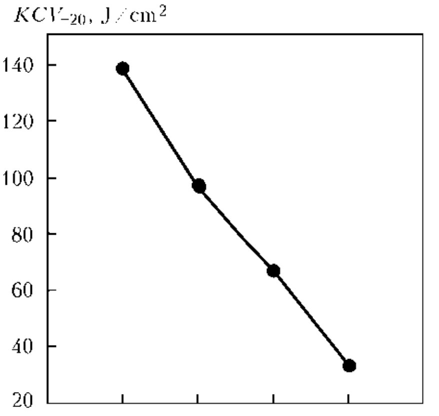


Figure 5. Dependence of average values of joint impact toughness on HT temperature

Lower limit of normalizing temperature was determined on the basis of dylatometric analysis of steel performed in Gleeble unit. According to dilatometric curves, at heating duration of 5 min complete austenitizing is achieved at temperature A_{c3} = 911 °C. Considering the experience of conducting the normalizing ensuring temperature rise by 30–40 °C above point A_{c3}, the lower temperature of normalizing was taken to be 950 °C.

At development of optimum HT mode normalizing temperatures in the range of 950–1100 °C were considered.

According to the results of testing the joints (Figure 5), average values of impact toughness KCV_{-20} decreased monotonically from 140 J/cm^2 at 950 °C to 38 J/cm^2 at 1100 °C with HT temperature rise.

Metallographic examination of the joint after HT showed that metal microstructure in the joint zone is a ferritic matrix with islet inclusions of residual austenite (Figure 6).

Ferritic matrix is granular and consists predominantly of polygonal mesoferrite, formation of which along the joint line, unlike the metal layers adjacent to the joint line, is caused by metal decarbonization in this section.

Structurally non-uniform sections with increased carbon content are observed in residual austenite (Figure 7). Evidently, residual austenite partially decomposes. The decomposition product is the structure described in [7–10] as granular bainite, which is a mechanical mixture of bainitic α-phase and carbides.

HT temperature increase leads to ferrite grains growing from 3–10 μm at 950 °C up to 10–30 μm at 1050 °C (see Figure 6, *a*–*c*). At 1100 °C temperature (Figure 6, *d*) growth of ferrite grains is somewhat suppressed. However, volume fraction of residual austenite in the matrix increases.

It is known [7] that increase of grain size and quantity of the second phase are exactly the factors, lowering the cold resistance. This accounts

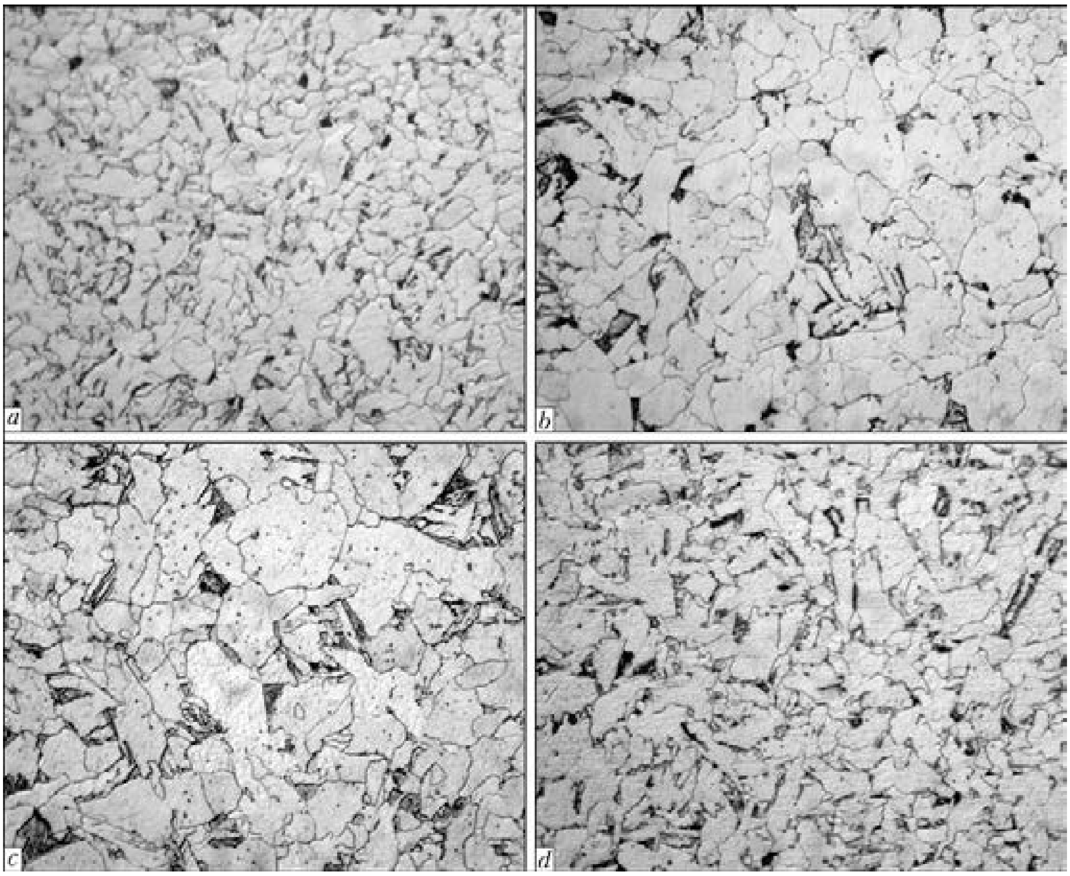
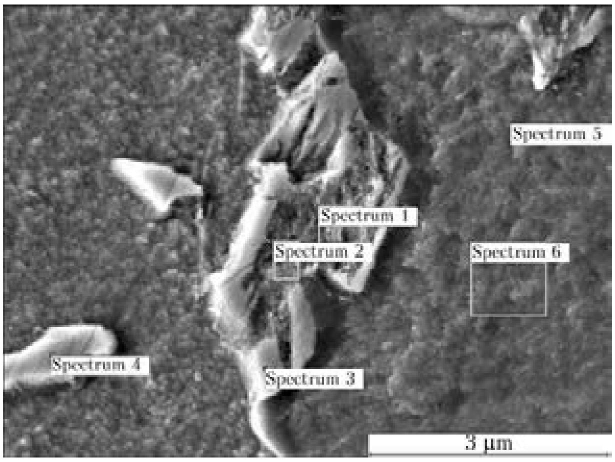


Figure 6. Microstructures (×1000) of metal in the zone of butt welded joints after HT at temperature of 950 (*a*), 1000 (*b*), 1050 (*c*) and 1100 (*d*) °C

for the shape of the curve of temperature dependence of impact toughness.



Spectrum number	C	O	V	Fe
1	2.37	2.23	0	95.41
2	3.20	2.35	0.86	93.59
3	1.81	1.33	0	96.86
4	1.48	1.27	0	97.25
5	1.01	0.88	0.43	97.68
6	1.24	1.13	0.13	97.50

Figure 7. Microstructure and chemical inhomogeneity of residual austenite, wt.%

1000 °C is the optimum HT temperature, allowing for the possible non-uniformity of the temperature field across the item thickness.

Figure 8 illustrates the influence of the time of sample soaking at normalizing temperature (1000 °C) on impact toughness value. Average value of impact toughness at 1 min soaking time was equal to 47 J/cm^2 (minimum values for individual samples do not exceed 30 J/cm^2).

Increase of soaking duration above 1 min leads to impact toughness increase. It is obvious that 1 min is insufficient for complete recrystallization of steel.

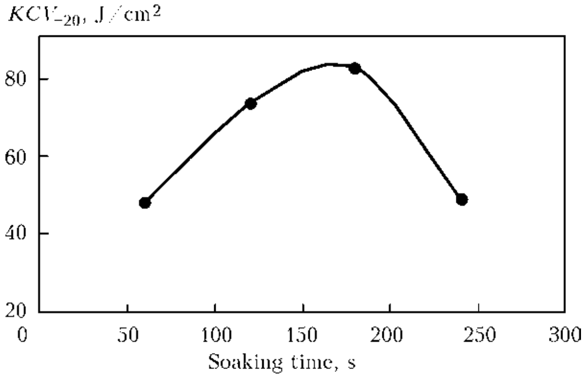


Figure 8. Dependence of average values of impact toughness on soaking at T_{HT} = 1000 °C

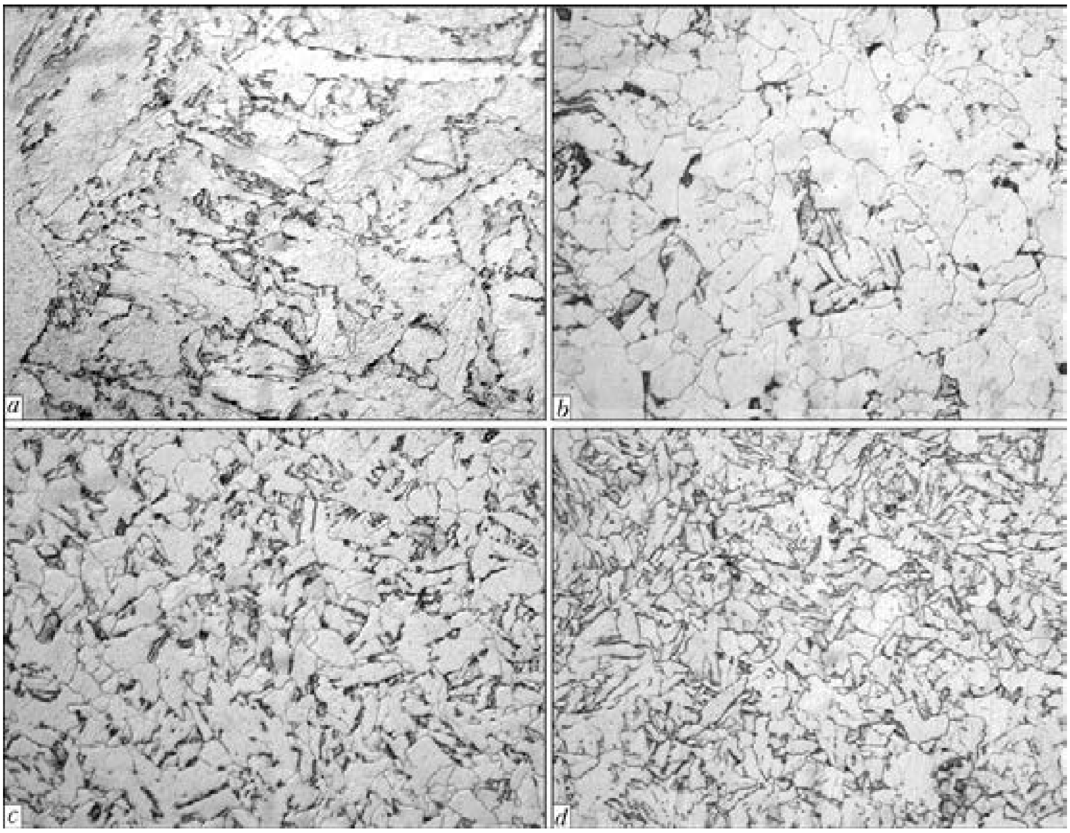


Figure 9. Microstructures (×1000) of metal in the zone of butt welded joints after HT at 1000 °C with soaking for 1 (*a*), 2 (*b*), 3 (*c*) and 4 (*d*) min

Dependence of impact toughness on soaking time is of an extreme nature. The highest and most stable KCV_{-20} values were obtained in the range of 2.5–3.0 min, and were equal to more than 80 J/cm^2. Increase of soaking duration above 3 min leads to impact toughness lowering and at 4 min soaking the impact toughness was equal to about 50 J/cm^2.

An abrupt lowering of impact toughness at soaking time longer than 3 min is due to a change in metal microstructure. Thus, while at 2 min soaking the matrix mostly consists of polygonal mesoferrite (Figure 9, *b*), at 4 min soaking feathery-acicular ferrite (Figure 9, *d*) prevails, which is identified as the product of shear transformation, unlike mesoferrite, forming by the diffusion mechanism. Residual austenite is located along ferrite needles.

Formation of such a microstructure occurs, probably, due to austenitic matrix enrichment in carbon and, consequently, inhibition of mesoferrite formation and increase of the fraction of residual austenite.

Joint line enrichment in carbon is possible as a result, on the one hand, of homogenizing of its concentration in the joint at HT (after welding the joint line is depleted in carbon) and, on the other hand, as a result of decomposition of thermally unstable vanadium carbides. Both the processes develop in time that is reflected on the curve of impact toughness dependence on HT duration.

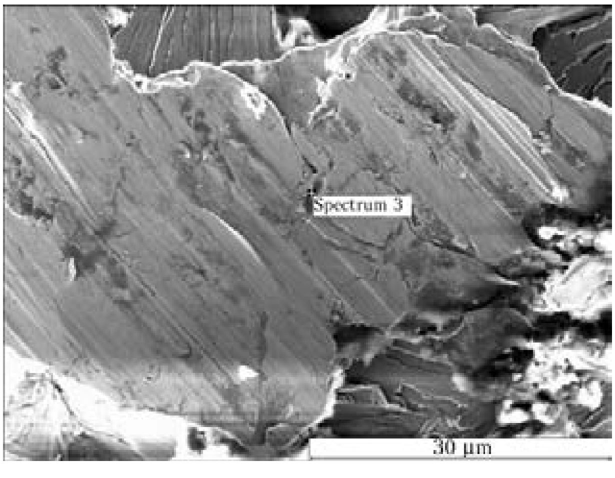


Figure 10. Fragment of joint fracture surface

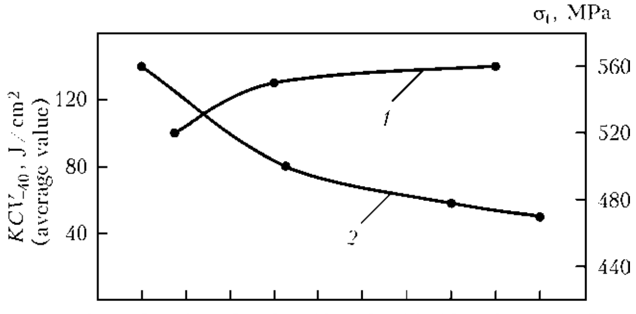


Figure 11. Dependence of tensile strength (*1*) and impact toughness (*2*) on cooling rate at HT

Indirect confirmation of diffusion-induced redistribution of carbon in the matrix was obtained at investigation of joint fractures. Fracture surface showed sections with high carbon content. In spectrum 3 (Figure 10) element content is as follows, at.%: 58.78 C; 5.79 O; 35.44 Fe.

Rate of cooling after HT has a dual influence on the joint properties: on the one hand, its increase leads to improvement of the joint strength properties and, on the other hand, toughness properties decrease (Figure 11). Optimum cooling rate, in terms of ensuring the set of mechanical properties, is in the range of 8–12 °C/s. At such cooling rates the indices of welded joint ultimate strength are preserved on the level of base metal strength.

Proceeding from analysis of the above data and allowing for the possible non-uniform heating of large-diameter pipes under the conditions of the actual construction, the following mode of pipe HT can be realized: heating temperature of 1000 °C, soaking time of 2.5–3.0 min, cooling by water-air mixture.

The above results of mechanical testing of the joints after basic HT show that they meet the requirements made of circumferential pipe joints [2].

Conclusions

1. Basic technology of HT of FBW joints on thick-walled pipes from steels of K56 grade was developed, which ensures compliance to the requirements of DNV-OS-F101 and API 1104 made of mechanical properties of welds for underwater pipelines.

2. Optimum temperature of welded joint normalizing is equal to 950–1000 °C, soaking duration at this temperature is equal to 2.5–3.0 min, cooling rate should be within 8–12 °C/s.

3. Rate of heating at HT does not have any essential influence on joint toughness values.

1. Morozov, Yu.D., Matrosov, M.Yu., Nastich, S.Yu. et al. (2008) High-strength pipe steels of new generation with ferrite-bainite structure. *Metallurg*, **8**, 39–42.
2. (2000) *DNV-OS-F101*: Offshore standard. Submarine pipeline systems. Jan. 2000.
3. Kuchuk-Yatsenko, S.I., Shvets, Yu.V., Zagadarchuk, V.F. et al. (2012) Flash-butt welding of thick-walled pipes from high-strength steels of K56 strength class. *The Paton Welding J.*, **5**, 2–7.
4. Pogorzhelsky, V.I., Litvinenko, V.I., Matrosov, Yu.I. et al. (1979) *Controlled rolling*. Moscow: Metallurgiya.
5. Heisterkamp, F., Hunna, K., Matrosov, Yu.I. et al. (1999) *Niobium-containing low-alloy steels*. Moscow: Intermet Engineering.
6. Livshits, L.S., Khakimov, A.N. (1989) *Metals science of welding and heat treatment of welded joints*. Moscow: Mashinostroenie.
7. Grabin, V.F., Denisenko, A.V. (1978) *Metals science of welding of low- and medium-alloy steels*. Kiev: Naukova Dumka.
8. Svishchenko, V.V. (1995) *Structure and mechanism of formation of granular bainite*: Transact. of I.I. Polzunov Altaj STU. Barnaul: AltGTU.
9. Kremnev, L.S., Svishchenko, V.V., Cheprasov, D.P. (1997) Structure and mechanism of granular bainite formation in 20Kh2NAch steel. *Metallovedenie i Termich. Obrab. Metallov*, **9**, 6–9.
10. Svishchenko, V.V., Cheprasov, D.P., Shabalin, V.N. et al. (2008) Formation of intermediate structure of granular morphology in low-carbon low-alloy steel under the conditions of step-isothermal decomposition of austenite. *Polzunovsky Almanakh*, **3**.

Received 29.11.2012