PROPERTIES OF THE WELD METAL OF TWO-SIDED WELDED JOINTS ON PIPES MADE FROM INCREASED-STRENGTH MICROALLOYED STEELS

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Two-pass two-sided welding is widely applied to manufacture large-diameter pipes for main gas and oil pipelines. It is obvious that metal of the welds that are the first to make is subjected to repeated heating in making the subsequent pass. The study is dedicated to evaluation of the effect of repeated heating on properties of the weld metal of the two-sided welded joints on pipes made from increased-strength microalloyed steels. Investigations were carried out on the weld metal of welded joints on the pipes made from ferritic-pearlitic steels of strength category K56–K60 and of different microalloying types (10(09)G2FB, 10G2FT, 10G2T, etc.). Impact toughness, hardness and peculiarities of structural characteristics of the inside weld metal of the welded joints on pipes subjected to repeated heating in making the outside weld were evaluated. It was shown that decrease in impact toughness of test specimens of the two-pass two-sided welded joints on the pipes, the tested section of which comprises the metal of intersection of the inside and outside welds, is caused by the presence of the local embrittlement zones forming in the inside weld metal due to its heating in making the outside weld. Investigations, including by transmission electron microscopy combined with microdiffraction, detected the following two zones present in the inside weld metal: low-temperature zone (heating to 450–650 °C) caused by occurrence of the dispersion hardening processes, and high-temperature zone (heating to 950–1100 °C) related to the MAC-phase containing a substantial amount of a more stressed lath martensite and forming as a result of decomposition of non-homogenised austenite. 8 Ref., 3 Tables, 6 Figures.

Keywords: *gas and oil line pipes, welded joint, weld metal, repeated heating, impact toughness, hardness, structure*

In case of two-pass two-sided welding, which is widely applied in lines of production of large-diameter pipes for main gas and oil pipelines, some regions of the earlier welded, first (inside) weld are subjected to heating to different temperatures in making the second (outside) weld. Naturally, this causes changes in structure and properties of metal in the local zones and in the welded joint as a whole.

In evaluation of impact toughness of the weld metal, the effect of such local zones is most pronounced on the pipes with a relatively small wall thickness (e.g. 12–18 mm), when section of a standard impact test specimen (10 × 10 mm) inevitably includes the metal of the first (inside) weld, in addition to the metal of the second (last) outside weld. However, for the pipes also with a bigger wall thickness, the modern specifications more and more often provide for additional tests of impact specimens cut out from metal of the first weld or from the location where the first and second welds intersect, which comprises regions subjected to repeated heating [1].

Decrease in toughness of the weld metal on microalloyed steel pipes subjected to repeated heating to temperatures of 450–650 °C is a well-known fact. Its nature and the effect of separate microalloying elements are well-studied [2–4]. It is generally recognised that embrittlement of such a weld metal is caused by the dispersion hardening process. The data on properties of metal of the welds subjected to heating to higher temperatures are extremely limited [5].

The purpose of this study was to investigate impact toughness, structural parameters and hardness of metal of the inside weld on the microalloyed steel pipes subjected to repeated heating in making the outside weld, depending on its chemical composition.

Analysed were the results of investigations of welds of the welded joints on pipes with a diameter of 1020–1420 mm and wall thickness of 15.7–30 mm, made from ferritic-pearlitic steels of strength category K56–K60 and of different microalloying types (10(09)G2FB, 10G2FT, 10G2T, etc.), as well as welds of the welded joints on steels of similar microalloying types, made under laboratory conditions. The welds were made by using wire of the Fe–Mn–Ni–Mo or Fe–Mn–Ni–Mo–Cr systems, and aluminate or

Table 1. Experimental compositions of welds (to Figure 1)

Code of welded joint	Steel/wire alloying type (steel thickness, mm)	Content of main alloying elements in weld metal, wt.%						
		C	Mn	Mo	V	Nb	Cr	Ti
3	Mn–V–Nb/Mn–Ni–Mo (18.7)	0.080	1.68	0.30	0.020	0.022	0.09	0.011
4	Mn–V–Nb/Mn–Ni–Mo–Cr (15.7)	0.080	1.75	0.20	0.060	0.033	0.24	0.020
6	Mn–V–Nb/Mn–Ni–Mo (30)	0.070	1.72	0.25	0.050	0.032	0.06	Traces

Table 2. Experimental compositions of welds (to Figure 2)

Code of welded joint	Steel/wire alloying type (steel thickness, mm)	Content of main alloying elements in weld metal, wt.%						
		C	Mn	Mo	V	Nb	Cr	Ti
5	Mn–Ti/Mn–Ni–Mo (15.7)	0.068	1.80	0.19	Traces	Traces	0.01	0.035
6	Mn–V–Nb/Mn–Ni–Mo (30)	0.070	1.72	0.25	0.050	0.032	0.06	Traces
7	Mn–V–Ti/Mn–Ni–Mo (17.5)	0.082	1.75	0.24	0.075	Traces	0.05	0.010

Table 3. Experimental compositions of welds (to Figure 3)

Code of welded joint	Steel/wire alloying type (steel thickness, mm)	Content of main alloying elements in weld metal, wt.%						
		C	Mn	Mo	V	Nb	Cr	Ti
5	Mn–Ti/Mn–Ni–Mo (15.7)	0.068	1.80	0.19	Traces	Traces	0.01	0.035
6	Mn–V–Nb/Mn–Ni–Mo (30)	0.070	1.72	0.25	0.050	0.032	0.06	Traces
8	Mn–V–Nb/Mn–Ni–Mo (17.5)	0.065	1.75	0.18	0.056	0.048	0.06	Same
9	Mn–V–Nb/Mn–Ni–Mo–Cr (15.7)	0.080	1.78	0.20	0.060	0.033	0.24	0.020

high-silicon fused flux (of the AN-60 or AN-67B type). Impact toughness KCV_{-20} of the weld metal of such joints ranges, mainly, from 30 to 100 J/cm^2.

As the investigations were carried out by using a rather large number of different variants of alloying of the welded joints, for convenience of description of the results the data on chemical composition of metal of the investigated welds are presented separately for each particular case (Tables 1–3 corresponding to Figures 1–3).

For impact bend tests of the two-sided welded joints on pipes, the specimens were cut out on the side of the weld that was the last to make (Figure 1, *a*; specimen H), as well as from the zone of intersection of the welds (specimen C). The tests often fixed decrease in impact toughness of the metal in a case where the section of a test specimen comprised a certain amount of the inside weld subjected to repeated heating. As an example, Figure 1 shows a characteristic change in the level of impact toughness of metal of the welds of different microalloying types depending on the amount of metal of the first weld contained in section of the test specimen. As follows from the data presented, the values of impact toughness decrease with increase in this amount, especially at a relatively high content of carbide-forming elements (Figure 1, *a*; welded joint 4, compared to welded joint 3). This difference in impact toughness increases with decrease in test temperature (Figure 1, *b*), the drop of average values of impact toughness amounting to 30 J/cm^2 (here and in other similar parts of the study the average values of KCV are the results of tests of three to six specimens).

Special impact bend tests were conducted on the weld metal of a characteristic chemical composition (with different contents of carbide-forming elements), at which a specimen was located in thickness of the welded joint in such a way that the notch bottom was at a different distance from the penetration line of the second weld, i.e. in regions of the first weld metal subjected to heating to different temperatures.

Figure 2 shows the level of impact toughness of metal of the welds of different chemical compositions depending on the location of the notch bottom in a specimen. It can be seen that in the immediate vicinity of the penetration line of the second weld the impact toughness is even a bit higher than the level characteristic of the weld metal not subjected to heating. As a rule, a minor decrease in hardness of the metal is fixed in this

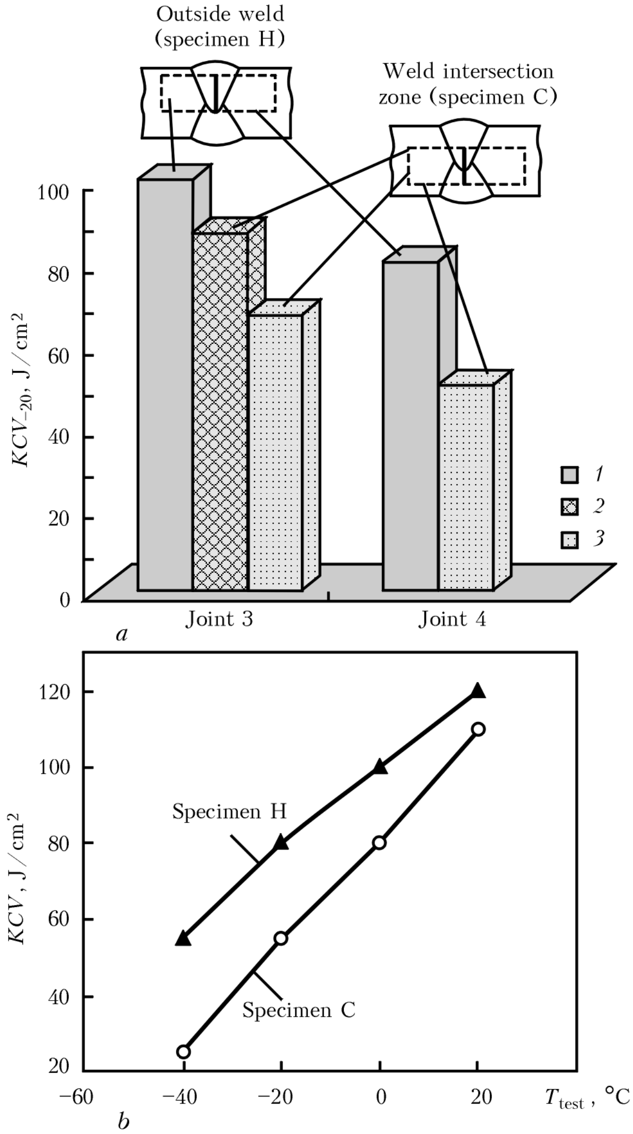

Figure 1. Impact toughness of metal of the welds at different amounts of the inside weld contained in section of test specimen (*a*), and effect of the test temperature on impact toughness value (*b*): *a* — welded joints 3 and 4; *b* — 6; *1* — 0; *2* — 15; *3* — 25 % of the inside weld

zone (Figure 3). With further distance of the notch bottom from the second weld penetration boundary the zones of decreased toughness are revealed in metal of the first weld. One of such zones is fixed at a distance of 1.5–3.0 mm from the penetration boundary. As the temperature of repeated heating of metal in this zone is 950–1100 °C, it can be classed with a conditionally high-temperature embrittlement zone (HTEZ).

The said embrittlement was found to occur in base metal of the welds with a relatively increased content of molybdenum and other carbide-forming elements, and, first of all, vanadium, niobium or chromium (e.g. welded joints 6–9 in Figures 2 and 3 with total content of V + + Nb + Cr + Ti + Mo = 0.38–0.56 wt.%). Impact toughness of the weld metal in this zone is 15–20 J/cm² lower than that of the second weld

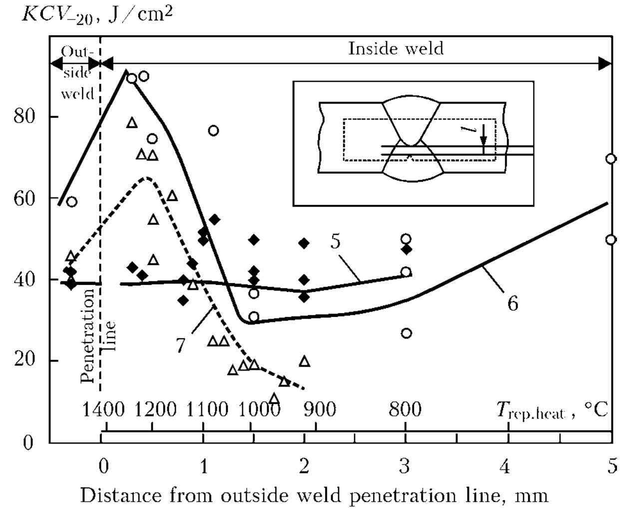

Figure 2. Effect of location of the notch on impact toughness of the investigated weld metal on pipes

metal not subjected to repeated heating, and hardness as a rule is higher by HV 25–35 (see Figures 2 and 3). In the weld metal alloyed only with molybdenum and titanium (e.g. welded joint 5 in Figure 2 with Ti + Mo = 0.225 %) no such decrease in impact toughness was fixed.

Another region of decrease in impact toughness of the first weld metal, where embrittlement develops because of dispersion hardening, is located at a distance of 4–6 mm from the second weld penetration boundary. The temperature of repeated heating of metal in this zone is 450–650 °C. Naturally, the degree of hardening (embrittlement) of metal in this case depends on the content of carbon and carbide-forming elements. Because, as indicated above, the nature of decrease in toughness of the weld metal due to the dispersion hardening processes is well studied, later on the HTEZ of the first weld metal was investigated in more detail.

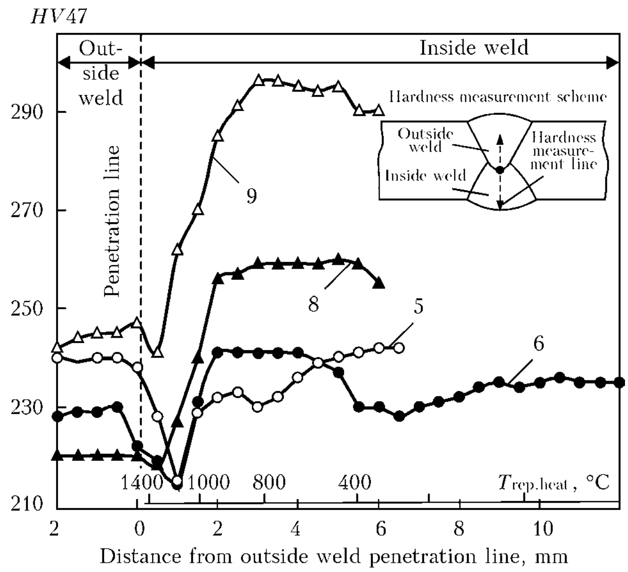

Figure 3. Distribution of hardness in the weld metals of different chemical compositions

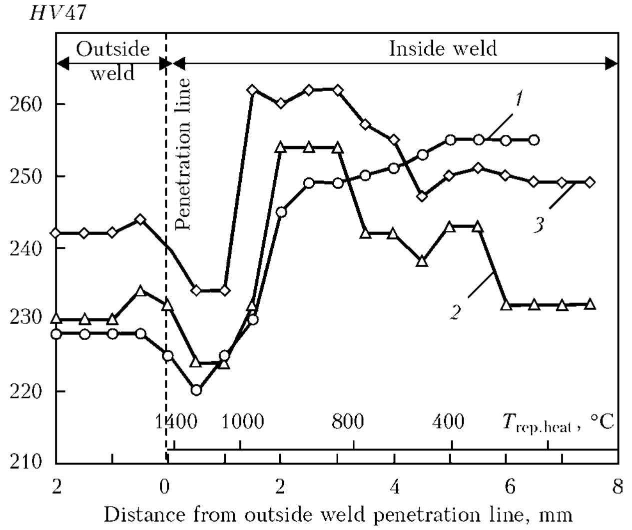

Figure 4. Effect of accelerated cooling on distribution of hardness in the weld metal: *1* — cooling in air; *2* — water cooling; *3* — water-air cooling

As shown by the additional tests, accelerated cooling during the welding process enhances embrittlement of the high-temperature zone. For example, in air, water and water-air mixture cooling of the first weld metal with a total content of vanadium, niobium, chromium, titanium and molybdenum at a level of 0.40 %, the maximum hardening of the HTEZ metal was fixed in a case of cooling by the water-air mixture, i.e. at a higher cooling rate (Figure 4). Also, it should be noted that in a case of cold crack resistance tests by the LTP2-6 procedure [6] providing for accelerated cooling, also an intensive hardening and even cracking were observed particularly in HTEZ (Figure 5, curves *1* and *2*) of the first weld metal with a chemical composition close to the indicated one (the total content of carbide-forming elements — at a level of 0.39 %). At the same time, under identical test conditions no hardening was fixed, and cracks were absent in the first weld metal alloyed only with molybdenum and titanium and close in chemical composition to the weld of welded joint 5 (the total content of carbide-forming elements — at a level of 0.23 %) (Figure 5, curve *3*).

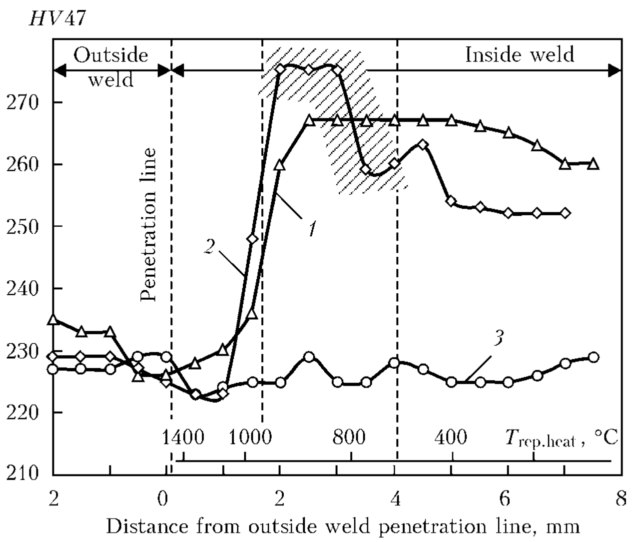

Figure 5. Distribution of hardness in the weld metal subjected to cold crack resistance tests: *1* — as-welded; *2*, *3* — after the cold crack resistance tests; *1*, *2* — weld of the Mo–V–Nb alloying system; *3* — weld of the Mo–Ti alloying system (dashed region — cold crack formation zone)

Optical metallography of probable changes in structure of the inside (first) weld metal as a result of repeated heating in making the outside (second) weld revealed no substantial differences in parameters of the structural state of metal of the local zones in the inside weld subjected to heating to different temperatures, where a marked increase in hardness of metal in the high-temperature zone and in the dispersion hardening zone had been fixed. Structural characteristics determining the level of impact toughness of the weld metal [7, 8] (contents of different structural components, i.e. acicular ferrite, intergranular polygonal or lamellar hypoeutectoid ferrite, upper bainite, and sizes of these formations, morphology and distribution of a microphase consisting of the martensite-austenite-carbide complexes (MAC-phase), quantity, size and morphology of non-metallic inclusions) were typical of the welding consumables and materials welded. The exceptions were the welds with a maximal content of molybdenum and niobium, where substantial formations of the carbon phases and structural components (MAC-phase, pearlite, carbides) were fixed in the said regions along the crystalline grain boundaries. Therefore, the local embrittlement zones were additionally examined by the transmission electron microscopy method by using foils, in a combination with microdiffraction).

Examinations were carried out mainly with the welds made by using flux AN-60 and wire Sv-08GNM on steels with a carbon content at a level of 0.1 %, and with different contents of microalloying elements and nitrogen. Such characteristics of the structural-phase state of the weld metal as peculiarities of acicular ferrite and dislocation structure, presence of other austenite transformation products, morphology and distribution of phase precipitates were evaluated.

The examinations showed that at a relatively low content of nitrogen (no more than 0.006 %) and carbide-forming elements (Ti + V + Nb = = 0.04 %, where Ti = 0.03 %) the acicular ferrite microstructure with a well-developed sub-structure (sizes of sub-structural elements are approximately (0.06–1.5) × 2(.0–7.0) μm, and form-factor χ is about 2–4) forms in metal of the inside weld, both in HTEZ and in the tempering zone. Dislocation density ρ is at a level of 10^{10} cm^{-2}. The structure is characterised by an ordered distribution of dislocations of volumetric equilibrium configurations. The boundaries of the sub-structural elements, having the form of dislocation networks and webs (inclination and torsion boundaries) are indicative of the relaxation processes occurring in the weld metal. In addition to acicular ferrite and a small amount of polygonal hypoeutectoid ferrite (10–15 %), also microregions of the intermediate transformation products (Figure 6, *a*), containing the indistinct fine-lamellar dispersed carbide inclusions (probably, in content of the MAC-phase), were revealed. Phase precipitates with size of about 0.006–0.012 μm were detected in the tempering zone in ferrite grains. The distance between separate particles of the phase precipitates is approximately 0.06–0.16 μm. The microdiffraction analysis of composition of the phase precipitates allowed identifying them as titanium, iron and vanadium carbides.

Increase in the content of carbide-forming elements in the inside weld metal (Ti + V + Nb = = 0.05 %, where there is almost no titanium), leads to formation of acicular ferrite, the grains of which are more elongated (χ = 3–5, and up to 10 for individual sub-grains) and similarly oriented in some areas. Sizes of the sub-structural elements are about (0.8–1.2) × (3.5–5.0) μm. The quantity of bainite microvolumes increases, and twinned martensite (in content of the MAC-phase) appears in the HTEZ region. The dislocation density ρ is at a level of 10^{10} cm^{-2}. The phase precipitates in the tempering zone, which are mostly vanadium, niobium and iron carbides, are 0.02–0.03 μm in size. The distance between the particles is 0.1–0.2 μm.

A simultaneous increase in the contents of nitrogen (up to 0.010–0.012 %) and microalloying elements (Ti + V + Nb > 0.06 %) leads to more substantial changes in the dislocation structure and in the distribution of the phase precipitates. In this case also a dispersed fragmented structure of acicular ferrite forms (χ = 3–5 on the average, and up to 11 for individual sub-grains). The structure is characterised by high dislocation density ρ — up to 10^{11} cm^{-2} or higher. The distribution of dislocations is mainly chaotic. More microvolumes with the martensite morphology are detected in the structure (Figure 6, *b*). A distinctive feature of structure of such a weld in the tempering zone is a very high density of phase precipitates and their dispersion degree. Sizes of the phase precipitates are about 0.003–0.007 μm. The distance between the individual particles does not exceed 0.007 μm, which is practically commensurable with sizes of the precipitates proper. The precipitates are mostly titanium, aluminium, niobium and vanadium nitrides (carbonitrides). Such a structural state leads to formation of a more stressed structure of the weld metal, this being related to development of the dislocation, sub-structural strengthening mechanisms, as well as strengthening by the Orowan mechanism (with finely dispersed particles).

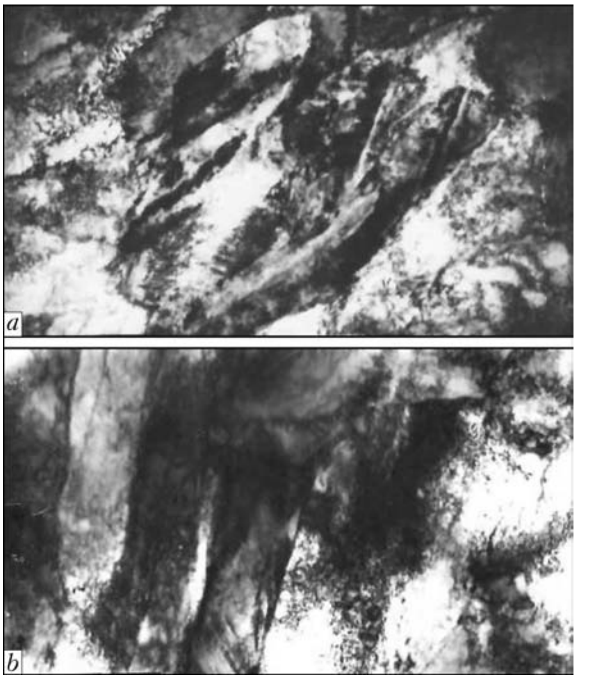

Figure 6. Microstructure (×15,000) of microphase in the HTEZ of metal of the inside weld at different alloying: *a* — Ti + V + Nb = 0.04 % (Ti = = 0.03); *b* — Ti + V + Nb > > 0.06 % (there is almost no titanium)

Therefore, the investigations confirmed formation of a large number of the dispersed (0.003–0.005 μm in size) particles of the VNb(CN) type located at a distance of about 0.007 μm from each other in the low-temperature hardening zone of the first weld, which is indicative of occurrence of the dispersion hardening process in the tempering zone.

For HTEZ, increase in the amount of the MAC-phase with the lath martensite microregions prevailing in its content was fixed. Molybdenum carbides and vanadium and niobium carbonitrides observed in this zone are relatively coarse, because the strengthening particles are inefficient.

Based on the data obtained, it seems possible to suggest the following version of the nature of HTEZ: it forms at a repeated heating temperature of about 950–1100 °C, when vanadium and niobium carbonitrides, as well as molybdenum carbides have dissolved in the weld metal. Under these conditions a not yet homogenised austenite in the local zones, especially at an increased level of alloying, decomposes in the process of subsequent cooling at decreased temperatures to form the microphase containing a more stressed lath martensite, in addition to the bainite transformation products, this increasing the sensitivity of this zone to embrittlement. Under certain conditions this may lead to formation of cold cracks in the welds.

Conclusions

1. Decrease in impact toughness when testing specimens of the two-pass two-sided welded joints on pipes, the tested section of which comprises the metal of intersection of the inside and outside welds, is caused by a number of factors, the decisive one of which is the presence of the local embrittlement zones forming in metal of the inside welds due to its heating in making the outside weld. The inside weld metal comprises two such zones: in the high- and low-temperature heating ranges.

2. Formation of the low-temperature embrittlement zone is caused by the process of dispersion hardening of metal in repeated heating to temperatures of about 450–650 °C, and its effect is proportional to increase in the content of carbon and carbide-forming elements.

3. Investigations, including by transmission electron microscopy combined with microdiffraction, suggested that decrease in toughness of the high-temperature embrittlement zone forming in repeated heating of the inside weld to temperatures of 950–1100 °C is caused by formation of unfavourable structural components, such as the MAC-phase containing a marked amount of a more stressed lath martensite, due to decomposition of non-homogenised austenite.

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