INFLUENCE OF DIFFUSIBLE HYDROGEN ON DELAYED CRACKING RESISTANCE OF HIGH-CARBON STEEL WELDED JOINTS

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Influence of diffusible hydrogen in deposited metal on the change of resistance to and nature of delayed fracture of HAZ metal in welded joints of high-strength wheel steel of grade 2 with 0.58 wt.% C was studied. «Pencil» method was used to determine diffusible hydrogen content in the deposited metal in CO_2 welding by PP-AN180MN flux-cored wire, the amount of which can vary in the range from 0.3 up to 2.2 $cm^3/100$ g. At testing by the implant method the influence of diffusible hydrogen on the change of critical stress values at delayed fracture of wheel steel joints was evaluated. Scanning electron microscopy methods were used to study the influence of diffusible hydrogen on the nature of HAZ metal fracture; characteristic fracture zones and structural component parameters were determined. It is established that at diffusible hydrogen content in the deposited metal on the level of 0.3 $cm^3/100$ g delayed fracture resistance of the joints is the highest, and depending on structural condition of metal of HAZ overheated zone critical fracture stresses are equal to $(0.35-0.45)\sigma_{0.2}$. Fracture occurs predominantly in the brittle mode along the boundaries and through the grain body, fraction of tough structural component is not more than 20 %. At increase of diffusible hydrogen content HAZ metal becomes brittle and delayed fracture resistance decreases. The most abrupt drop of critical stress characteristics to $0.1\sigma_{0.2}$ value is characteristic for the metal of HAZ with martensite-bainite structure, which has 70 % of martensite. 13 Ref., 2 Tables, 6 Figures.

Keywords: *arc welding, diffusible hydrogen, wheel steel, delayed fracture, HAZ, structure*

Cold cracking in high-strength steel welded joints is known to proceed by delayed fracture mechanism. This requires presence of quenched metal structure (martensite, bainite) with a high level of inner stresses and diffusion-mobile hydrogen which, accumulating in structural discontinuities, increases local stress level. This way, diffusible hydrogen leads to further embrittlement of the structure, and initiates formation of microcracks in the metal. Further on under the impact of residual stresses formed in the joint in welding, microcracks develop into macrocracks [1–5]. Here, geometrical stress raisers in the joints such as undercuts, lacks-of-fusion, lacks-of-penetration in the root have a great influence, being the sites of residual stress localizing, where the probability of cold cracking is the highest [6].

It is obvious that hydrogen is one of the main factors in cold cracking in welded joints of high-strength steels. Hydrogen influence on the process of delayed fracture of metal is accounted for by dislocation model, which is described in detail in [5, 7, 8]. According to this model, atomic hydrogen present in the welded joint accumulates in propagating dislocations and lowers ductile properties of metal, energy of crack initiation and propagation at its fracture under the impact of external loading. It is also established that hydrogen influence becomes stronger with increase of metal strength.

Saturation of molten weld metal by atomic hydrogen, change of its solubility in the metal depending on temperature are described in [1]. So, hydrogen solubility in γ-iron is approximately equal to 8.2 $cm^3/100$ g at the temperature of 1200 °C, and in α-iron at 500 °C this characteristic is only 0.6 $cm^3/100$ g. Therefore, when at joint cooling structural $\gamma \to \alpha$ transformations proceed in weld metal, hydrogen solubility abruptly decreases and it diffuses into the ambient medium, including the adjacent region of HAZ metal. It is impossible to experimentally determine diffusible hydrogen content in HAZ metal.

In this connection, most of the researchers consider diffusible hydrogen content in deposited metal for comparative assessment of hydrogen influence on cold cracking resistance of high-strength steel welded joints. «Pencil» and chromatography methods are most often used for this purpose [1–3, 9]. For instance, it was established with application of «pencil» method that, depending on moisture level in welding consumables, diffusible hydrogen content in deposited metal in welding with solid wires in shielding

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gases can reach 3, in manual arc welding 12 and in submerged-arc welding 8 $cm^3/100$ g [1, 2].

Hydrogen content in HAZ metal can be assessed with a certain degree of approximation with application of calculation methods, allowing for kinetics of structural transformations in the welded joint. In [10] it was established that at initial content of diffusible hydrogen in the weld metal on the level of 10 $cm^3/100$ g, local amount of atomic (diffusible) and molecular (residual) hydrogen in the HAZ metal of joints of high-strength low-alloyed (HSLA) steel at cooling by the specified thermal cycle to 60 °C (welding heat input of 9.1 kJ/cm) can be within 1.09–1.59 and 0.38–1.68 $cm^3/100$ g, respectively. Changes of its concentration within the above limits depend on structural state of metal.

At formation of martensite-bainite structure content of diffusible and residual hydrogen in the HAZ metal is equal to 1.30 and 0.79 $cm^3/100$ g, and in bainite-ferrite structure it is 1.49 and 1.68 $cm^3/100$ g, respectively. As is seen, calculated local hydrogen concentrations in HAZ metal of the joints differ only slightly at its fixed content in the weld metal and constant welding heat input. Here diffusible hydrogen content in the HAZ metal is approximately 7 times lower than its initial content in the weld metal. However, even with such low and approximately same amount of it in the HAZ metal, the structure and ductile properties of which differ depending on TDWC conditions, delayed cracking resistance of HSLA steels is different.

In the case of arc welding of high-strength wheel steels, carbon content in which is higher than 0.5 %, quenching structures with higher dislocation density form in the HAZ metal. As a result, strength properties of the metal become higher, and ductile properties decrease [11–13]. It can be anticipated that in this case even negligible local hydrogen concentrations will lead to more essential changes of HAZ metal properties than in welding of HSLA steels. However, as investigations of weldability of high-strength steels with more than 0.5 % C began to be conducted comparatively recently, experimental data on hydrogen influence on delayed cracking resistance of welded joints of this type of steels are not available.

The objective of this work was investigation of the influence of diffusion-mobile hydrogen content in the deposited metal on the change of resistance to and nature of delayed fracture in HAZ metal of welded joints of high-strength wheel steel under static loading. Used as research material was wheel steel of grade 2 (GOST 10791) of the following composition, wt.%: 0.58 C; 0.44 Si; 0.77 Mn; 0.10 Ni; 0.05 Cr; 0.012 S; 0.011 P.

Quantitative evaluation of delayed fracture resistance of HAZ metal was performed with application of Implant method [2, 3]. Unlike the traditional method, samples-inserts from the studied steel of 6 mm diameter were made without notches [13]. HSLA steel 20 mm thick was used as technological plates. Samples after welding and their natural cooling to 50 °C were loaded by constant load. Comparative testing was performed with application of mechanized CO_2 welding by experimental 2 mm flux-cored wire PP-AN180MN (weld metal alloying of 10KhGSFT system) at welding current $I_w = 220$–250 A and arc voltage $U_a = 26$ V. Heat input at welding speed $v_w = 16.7$ m/h was $Q_w = 8.9$ kJ/cm, and at $v_w = 11$ m/h it was 13.6 kJ/cm. Welding of Implant samples was performed without preheating ($T_{pr} = 20$ °C). Here, HAZ metal cooling rate in the temperature range of 600–500 °C ($w_{6/5}$) was equal to 25–30 and 12–14 °C/s, respectively, and time of cooling from 800 to 100 °C ($\tau_{8/1}$) was 170 and 260 s, respectively.

It is obvious that metal structure in HAZ overheated zone will change depending on welding heat input and joint cooling rate. Specially conducted metallographic investigations showed that structure of wheel steel of grade 2 in as-delivered condition is represented by pearlite-ferrite mixture (Figure 1, *a*) with grain size $D_{gr} = 20$–75 μm and microhardness HV0.1-1990–2450 MPa. Ferrite fringes of size $h_f = 5$–10 μm are located along the grain boundaries. Under the conditions of welding, when cooling rate is equal to $w_{6/5} = 25$–30 °C/s, a mixed martensite-bainite structure forms in the HAZ metal overheated zone (Figure 1, *b*), in which the amount of martensite is 70 %, and structural component microhardness is HV0.1-4340–6990 MPa. At $w_{6/5} = 12$–14 °C/s martensite-bainite structure also forms in the HAZ metal overheated zone (Figure 1, *c*), but with prevailing fraction of bainite (80 %) with structural component microhardness HV0.1-3340–3680 MPa. Grain size in HAZ metal overheated zone is $D_{gr} = 63$–94 μm (point 4 and 5 to GOST 5639).

At testing by Implant method, hydrogen content in the deposited metal was varied by changing the temperature of flux-cored wire baking and time of its soaking in air before welding. Amount of diffusible hydrogen $[H]_{dif}$ in the deposited metal was determined by «pencil» method, using a mixture of glycerin and distilled

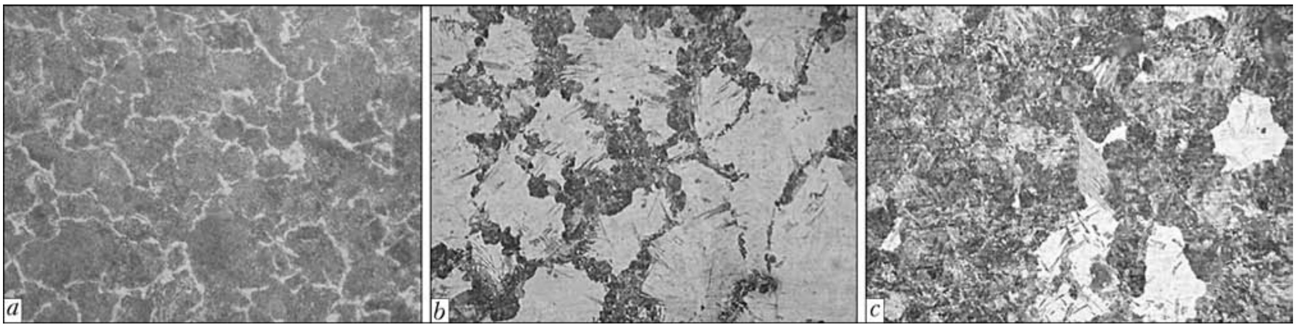

Figure 1. Microstructure (×500, reduced 2 times) of HAZ metal of grade 2 wheel steel: *a* — base metal; *b* — $w_{6/5}$ = = 25–30; *c* — 12–14 °C/s

water as blocking liquid. $[H]_{dif}$ content in the deposited metal depending on the conditions of preparation of flux-cored wire PP-AN180MN before welding was as follows, $cm^3/100$ g: baking at 230 °C for 2.5 h — 0.3; baking at 230 °C for 1 h — 0.5; baking at 230 °C for 1 h, seasoning in air for 7 days — 1; baking at 230 °C for 1 h, seasoning in air for 14 days — 1.3; baking at 200 °C for 2.5 h — 1.5; baking at 150 °C for 2.5 h — 1.8; without baking — 2.2.

It follows from the above data that in flux-cored wire CO_2 welding of high-strength carbon steels joints, it is possible to achieve a lower content of diffusible hydrogen in the weld metal, which is equal to 0.3 $cm^3/100$ g under the condition of pre-baking of the wire at 230 °C for 2.5 h. This value $[H]_{dif}$ is several times lower than in submerged-arc welding with solid wires. In welding with flux-cored wire, which was baked for 1 h, and its seasoning for 14 days in air (moisture level of up to 75 %), $[H]_{dif}$ in weld metal rises up to 1.3 $cm^3/100$ g. At lowering of baking temperature from 230 to 150 °C and in welding with wire not subjected to heat treatment, diffusible hydrogen content in the deposited metal is the highest — 1.8–2.2 $cm^3/100$ g.

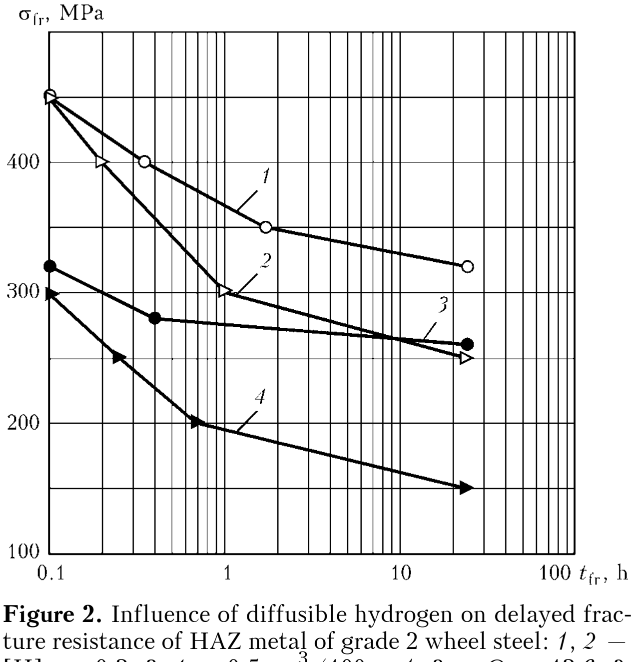

Figure 2. Influence of diffusible hydrogen on delayed fracture resistance of HAZ metal of grade 2 wheel steel: *1*, *2* — $[H]_{dif}$ = 0.3; *3*, *4* — 0.5 $cm^3/100$ g; *1*, *3* — Q_w = 13.6; *2*, *4* — 8.9 kJ/cm

Figure 2 gives data on the change of delayed fracture resistance of HAZ metal in joints of wheel steel of grade 2, depending on welding heat input and diffusible hydrogen content in the deposited metal, varying from 0.3 up to 0.5 $cm^3/100$ g. As is seen from the above material, critical stresses σ_{cr}, at which Implant samples do not fail, are the highest at minimum $[H]_{dif}$ values. In welding at heat input Q_w = 8.9 kJ/cm ($w_{6/5}$ = 25–30 °C/s), they are equal to 250 MPa, and at Q_w = 13.6 kJ/cm ($w_{6/5}$ = 12–14 °C/s) σ_{cr} rise up to 320 MPa.

Change of delayed fracture resistance values at constant $[H]_{dif}$ = 0.3 cm/100 g is influenced by structural state of metal in HAZ overheated zone. As was shown above, in welding of wheel steel in the modes, when Q_w = 13.6 kJ/cm, martensite-bainite structure of lower hardness forms in HAZ overheated zone. Amount of martensite in the structure is not higher than 20 %. Metal becomes more ductile, has better resistance to microcrack initiation and propagation and critical fracture stresses rise by 30 %.

It should be also noted that at Q_w = = 13.6 kJ/cm HAZ metal stays longer in temperature interval of 800–100 °C at cooling ($\tau_{8/1}$ is 1.5 times higher) than at Q_w = 8.9 kJ/cm. This can also promote increase of delayed fracture resistance due to development of hydrogen diffusion processes in HAZ metal.

At increase of diffusible hydrogen content in the deposited metal up to 0.5 $cm^3/100$ g delayed fracture resistance of wheel steel HAZ metal decreases. This occurs more intensively in the case of welding at Q_w = 8.9 kJ/cm ($w_{6/5}$ = 25–30 °C/s), when 70 % of martensite forms in the HAZ metal structure. Critical fracture stresses decrease from 250 to 150 MPa (by 40 %). In welding with a higher heat input (13.6 kJ/cm) and formation of predominantly bainitic structure in HAZ metal overheated zone, σ_{cr} decrease by 20 % to the level of 260 MPa.

As is seen, a slight increase of diffusible hydrogen in the deposited metal led to an essential lowering of the indices of delayed fracture resistance of HAZ metal. As shown in [13], in welding of joints of wheel steel with carbon content of 0.58 %, a structure, in which dislocation density can reach values $\rho = 5\text{–}8\cdot10^{10}$ cm^{-2}, forms in the HAZ metal overheated zone. Therefore, a comparatively small excess of the content of atomic hydrogen, which accumulates in dislocations with such density level in higher strength metal, leads to marked embrittlement of HAZ metal. This is indicated by comparative testing of Implant samples, in which welding heat input was equal to 8.9 kJ/cm at load level of 300 MPa (see Figure 1): time of HAZ metal fracture at $[H]_{dif} = 0.3$ and 0.5 cm^3/100 g in the weld metal is equal to 60 and 6 min, respectively.

Generalized results of investigation of the influence of diffusible hydrogen content in the deposited metal on delayed fracture resistance values of HAZ metal in welded joints of wheel steel of grade 2 (0.58 % C) are given in Figure 3. As is seen, subsequent increase of $[H]_{dif}$ content in the weld up to 1.0–1.5 cm^3/100 g leads to lowering of delayed fracture resistance of HAZ metal, but these changes proceed smoothly. So, at increase of diffusible hydrogen in the deposited metal by 1.0 cm^3/100 g (from 0.5 up to 1.5), σ_{cr} values for HAZ metal of wheel steel joints, made in the modes with $Q_w = 13.6$ kJ/cm, decrease by approximately another 12 % (from 260 to 220 MPa), and in welding with $Q_w = 8.9$ kJ/cm — by 24 % (from 150 to 90 MPa). Here also the influence of structural factor is manifested. Embrittlement of more hardened HAZ metal at increase of diffusible hydrogen proceeds to a greater degree.

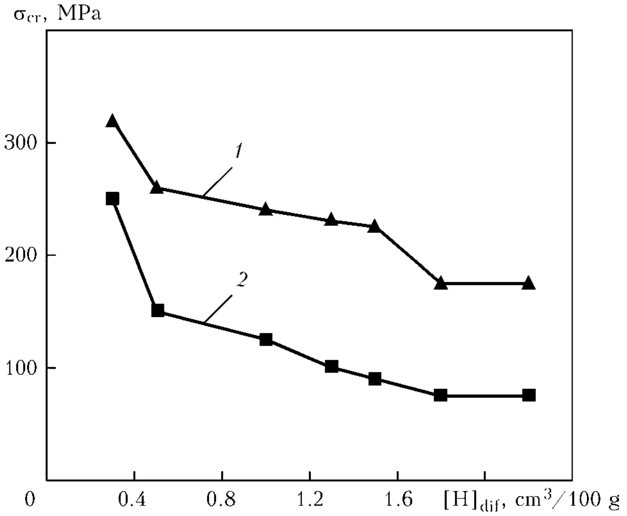

Figure 3. Influence of diffusible hydrogen content in the deposited metal on critical fracture stress of HAZ metal in joints of grade 2 wheel steel in welding with PP-AN180MN wire: *1* — $Q_w = 13.6$; *2* — 8.9 kJ/cm

At increase of $[H]_{dif}$ up to 1.8 cm^3/100 g delayed fracture resistance of HAZ metal of wheel steel joints, which were welded at $Q_w = 8.9$ kJ/cm, decreases by 6 % to values $\sigma_{cr} = 75$ MPa. This is the limit value of critical stresses under these conditions of welding, which is equal to approximately 0.1 of HAZ metal yield point ($\sigma_{0.2}$ of about 715 MPa [12]), and further on at increase of diffusible hydrogen content in the deposited metal to 2.2 cm^3/100 g, this value does not change. Total lowering of critical fracture stresses of HAZ metal in wheel steel joints at $Q_w = 8.9$ kJ/cm at increase of $[H]_{dif}$ content from 0.3 up to 2.2 cm^3/100 g in the deposited metal was equal to 70 %.

For HAZ metal of joints welded at $Q_w = 13.6$ kJ/cm, σ_{cr} value changes in a somewhat different manner at $[H]_{dif}$ increase from 1.5 up to 1.8 cm^3/100 g. As is seen from Figure 3, lowering of critical fracture stresses is equal to 14 % (from 220 to 175 MPa). This is, probably, related to the fact that under these welding conditions, when predominantly bainitic structure forms in the overheated zone of HAZ metal in wheel steel joints, its oversaturation with atomic

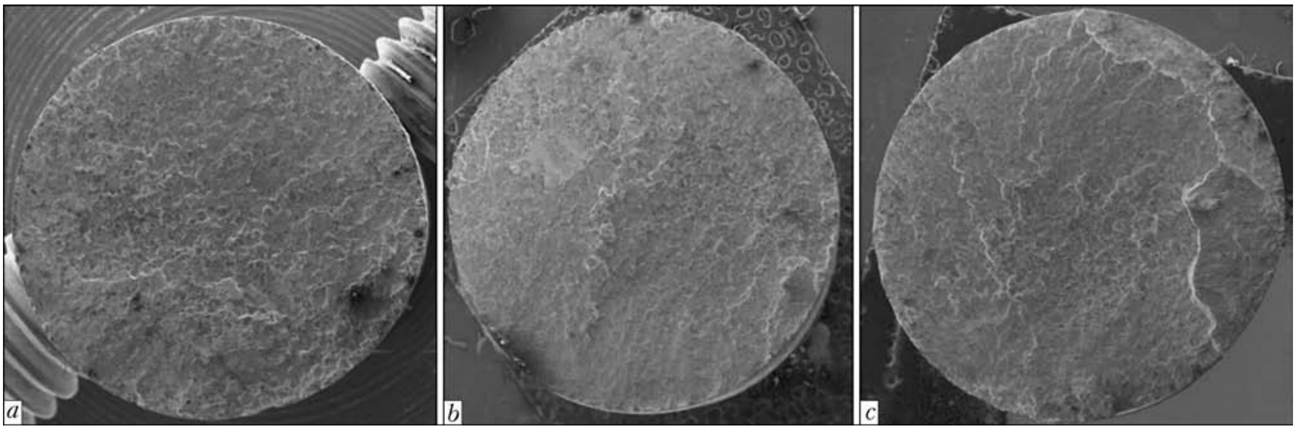

Figure 4. Fractures (×25) of Implant samples of grade 2 wheel steel in welding with PP-AN180MN wire at $Q_w = 13.6$ kJ/cm: *a* — $[H]_{dif} = 0.3$; *b* — 0.5; *c* — 2.2 cm^3/100 g

Table 1. Parameters of testing Implant samples of wheel steel for investigations of fracture surface (Q_w = 13.6 kJ/cm)

$[H]_{dif}$, cm^3/100 g	σ_{cr}, MPa	σ_{fr}, MPa	t_{fr}, h
0.3	320	350	1.5
0.5	260	320	0.1
1.5	220	250	4.5
2.2	175	200	4.5

hydrogen takes place and ductile properties of the metal decrease more abruptly. At further increase of diffusible hydrogen content in the deposited metal to 2.2 cm^3/100 g this value does not change, either. Total lowering of critical stresses of HAZ metal fracture in wheel steel joints at increase of $[H]_{dif}$ content from 0.3 up to 2.2 cm^3/100 g was equal to 45 % for these welding conditions.

Fractures of Implant samples of wheel steel, which were welded at Q_w = 13.6 kJ/cm with different content of diffusible hydrogen in the deposited metal, were studied by scanning electron microscopy methods*.

The Philips scanning electron microscope SEM-515, fitted with energy-dispersive spectrometer of LINK system, was used. Loading parameters at testing of Implant samples of wheel steel, which were selected for fracture surface studies, are given in Table 1.

Fractures of Implant samples of wheel steel of grade 2 are shown in Figure 4. It is established that fracture runs, mainly, in the brittle mode. Characteristic fracture zones are observed on

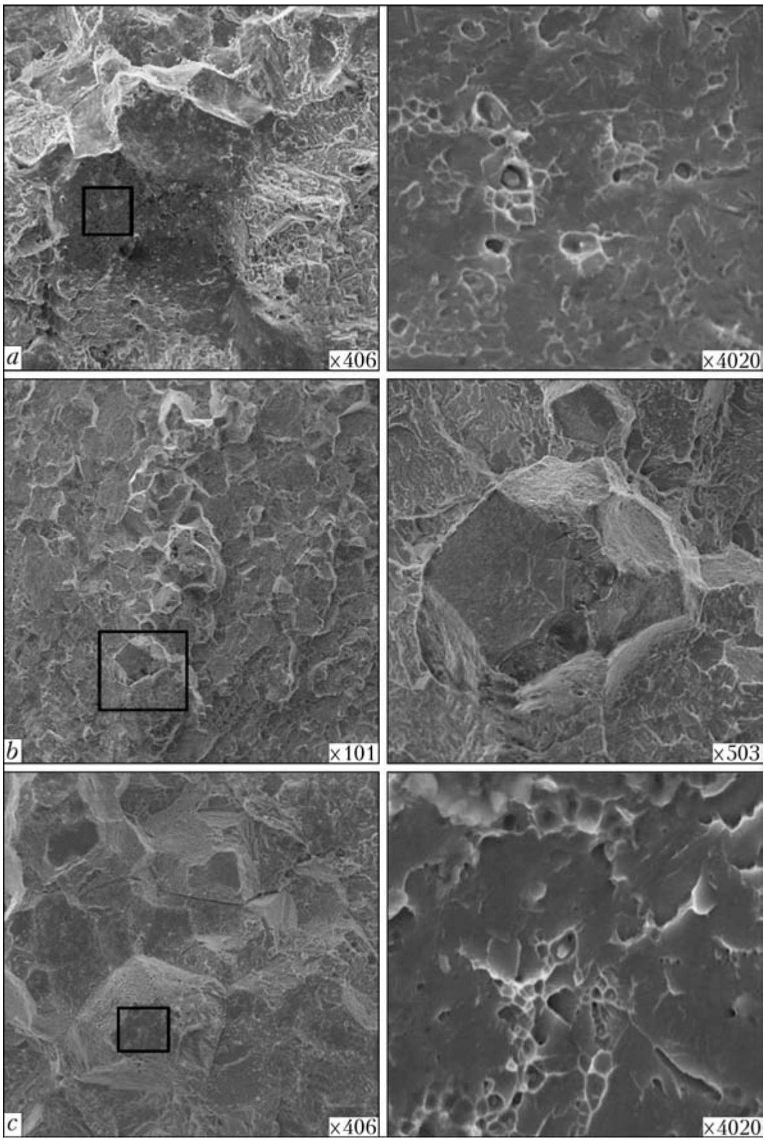

Figure 5. Fragments of fracture surface of HAZ metal in the region of initiation and delayed propagation of microcracks: *a* — $[H]_{dif}$ = 0.3; *b* — 0.5; *c* — 2.2 cm^3/100 g

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*Investigations were performed by L.I. Markashova, E.N. Berdnikova and T.A. Alekseenko.

Table 2. Fracture structure at delayed fracture of HAZ metal of grade 2 wheel steel

$[H]_{dif}$, $cm^3/100$ g	Zone I				Zone II					
	Brittle intergranular		Brittle intragranular		Brittle intergranular		Brittle intragranular		Tough	
	V, %	D_f, μm	V, %	D_f, μm	V, %	D_f, μm	V, %	D_f, μm	V, %	D_f, μm
0.3	20	30/100	80	50/100	20	60/100	60	40/60	20	0.5/2
0.5	40	50/120	60	50/100	15	50/120	80	40/60	5	0.5/2
1.5	60	50/120	40	50/100	40	50/100	60	50/80	–	–
2.2	80	50/100	20	50/80	40	50/100	60	40/80	–	–

fracture surface, which can be conditionally divided into local regions of microcrack initiation and their delayed propagation (zone I), regions of accelerated microcrack propagation up to macrolevel (zone II), and final fracture regions (zone III). Zone distribution is of a local nature, crack initiation and propagation occurs simultaneously in several sites, and fracture regions alternate. Investigations at microlevel showed that, depending on diffusible hydrogen content in the deposited metal and loading level, microrelief of HAZ metal fracture surface is different. Generalized results of investigations are given in Figures 5, 6 and Table 2.

It is established that microcrack initiation in the HAZ metal of wheel steel joints proceeds in

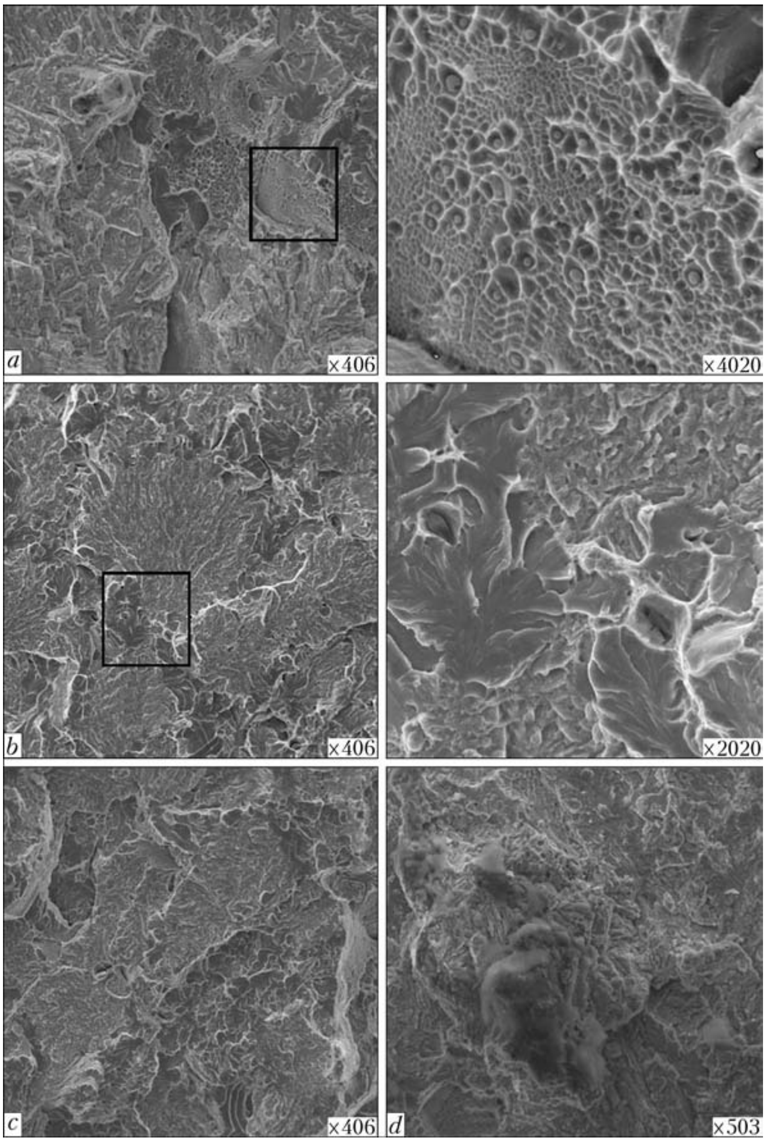

Figure 6. Fragments of fracture surface of HAZ metal in regions of accelerated development of microcracks (*a*, *b*) and final fracture (*c*, *d*): *a*, *c* – $[H]_{dif} = 0.3$; *b*, *d* – 2.2 $cm^3/100$ g

the brittle mode along grain boundaries. In fracture zone I with all welding variants, local regions of brittle intergranular fracture are observed, which are the sites of microcrack initiation (Figure 5, *a*–*c*). Further on, under the impact of stresses, microcracks propagate both along the boundaries and through the grain body: fracture in this zone is characterized as intergranular and intragranular cleavage. Size of cleavage facets D_f is equal to approximately 30–100 μm. Depending on diffusible hydrogen content in the deposited metal, ratio of these kinds of fracture changes. With increase of $[H]_{dif}$ content from 0.3 up to 2.2 cm^3/100 g, volume fraction V of intergranular cleavage rises from 20 up to 80 %, while fraction of intragranular cleavage decreases, respectively (see Table 3). In this zone also secondary cracks were detected, the length of which is equal to 250 μm at $[H]_{dif}$ = = 2.2 cm^3/100 g.

After the crack has reached its critical size, accelerated development of microcracks into macrocracks begins. Alongside intergranular and intragranular cleavage, local regions of tough fracture with facet size of about 0.5–2.0 μm are observed along the boundaries of some grains (Figure 6, *a*). At $[H]_{dif}$ = 0.3 cm^3/100 g volume fraction of tough fracture is up to 20 %, at $[H]_{dif}$ = = 0.5 cm^3/100 g it is 5 %, and at subsequent increase of diffusible hydrogen content in the deposited metal accelerated crack development runs solely in the brittle mode (Figure 6, *b*).

In the final fracture zone, fracture mode and size of structural elements differ from those of other fracture zones. This predominantly is intragranular quiasibrittle fracture with facet size of approximately 10–20 μm, and at increase of $[H]_{dif}$ content fracture mode practically does not change (Figure 6, *c*, *d*).

Thus, conducted investigations showed that:

- in welding with flux-cored wire of PP-AN180MN type, it is possible to ensure lower (0.3 cm^3/100 g) content of diffusible hydrogen in the deposited metal («pencil» method) that is much lower than in welding with solid wires in shielding gases and under a layer of flux. At such $[H]_{dif}$ value delayed fracture resistance of HAZ metal in joints of high-strength wheel steel with carbon content of 0.58 % is the highest. Critical fracture stresses, depending on structural state of the metal of HAZ overheated zone, were equal to approximately (0.35–0.45)$\sigma_{0.2}$. Fracture occurs predominantly in the brittle mode along the boundaries and in the body of grains, with not more than 20 % fraction of tough structural component;
- at increase of diffusible hydrogen content in the deposited metal by 0.2 cm^3/100 g (from 0.3 to 0.5) HAZ metal becomes brittle, and delayed fracture resistance of welded joints drops abruptly. Degree of this influence is determined by structural state of metal in HAZ overheated zone. At formation of martensite-bainite structure of higher strength and lower ductility (70 % of martensite, $w_{6/5}$ = 25–30 °C/s) critical fracture stresses decrease by 40 %, when a predominantly bainitic structure forms (80 % of bainite, $w_{6/5}$ = 12–14 °C/s) they decrease by 20 %. Here the fraction of tough component in the fracture is not higher than 5 %;
- at increase of diffusible hydrogen content in the deposited metal up to 1.8–2.2 cm^3/100 g further embrittlement of HAZ metal proceeds and critical fracture stresses decrease to minimum values of (0.10–0.25)$\sigma_{0.2}$. Fracture runs solely in the brittle mode along the boundaries and in the body of grains, fraction of brittle intergranular cleavage in the zone of microcrack initiation and delayed propagation rises from 20 up to 80 %.

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