# RESISTANCE TO COLD CRACK FORMATION OF HAZ METAL OF WELDED JOINT ON HIGH-STRENGTH CARBON STEELS

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Influence of preheating temperature and heat input during surfacing on change of delayed fracture resistance of HAZ metal on high-strength steel at variation of carbon content in it from 0.55 to 0.75 % was investigated. Effect of carbon content in steel on indices of critical stresses at HAZ metal delayed fracture was evaluated by Implant tests. Influence of a cooling rate on nature of metal fracture and typical rupture zone and parameters of structural constituents were studied and determined using the methods of scanning electron microscopy. It is determined that HAZ metal is predisposed to the delayed fracture in electric arc surfacing due to formation of quenched structures with high dislocation density and internal stresses in area of overheating. Increase of cooling rate and carbon content in steel promotes rise of fracture susceptibility and decrease level of critical stresses to $0.07\sigma_{0.2}$. Formation of more plastic structures at reduction of the cooling rate in 600–500 °C temperature interval promotes increase of the delayed fracture resistance of HAZ metal on high-strength carbon steels to $\sigma_{cr} \geq 0.45\sigma_{0.2}$ level. A diagram of effect of carbon content in steel on HAZ metal resistance to the delayed fracture was plotted in form of $w_{6/5} = f(C)$. It is determined that the process of delayed fracture in HAZ metal on steel with carbon content not more than 0.60 % can be prevented at cooling rate $w_{6/5}$ not more than 16 °C/s, $w_{6/5} \leq 8$ °C/s with 0.60–0.65 % C and $w_{6/5} \leq$ $\leq 5$ °C/s with 0.65–0.75 % C. Under such conditions, the structures, having sufficiently high capability to microplastic strain without microcrack generation, are formed in the metal of HAZ overheating area. 18 Ref., 2 Tables, 11 Figures.

***Keywords:*** *electric arc surfacing, high-strength carbon steel, HAZ, preheating, heat input, delayed fracture, structure, rupture*

Study of a phenomenon of delayed fracture of as-quenched metal under effect of constant loading was started as far back as the middle of the last century applicable to the conditions of heat treatment of quenching steels for providing of strength and service properties of the parts. Works [1–3] give the most complete explanation of physics of the delayed fracture process of quenched steel at loads, which are significantly lower than yield strength for given metal. It is also noted that the delayed fracture has exclusively brittle nature. In contrast to a classic brittle fracture, which takes place in short periods of time, the delayed fracture develops in the metal in a course of long period of time. At that, the processes of microplastic strain are developed in the local metal areas.

In welding the process of delayed fracture is realized at formation of cold cracks in the joints [4–6]. It is a well-known fact that the welded joints on high-strength steels are more susceptible to cold crack nucleation at formation in them of bainite-martensite structure, and increase of the carbon content in metal rises this susceptibility. Besides, saturation of weld metal with hydrogen, which diffuses in HAZ metal, takes palce in the process of welding. Presence of hydrogen intensifies the process of delayed fracture of the welded joints. It becomes more defined and receives own peculiarities.

Number of researchers [7–13] devoted their works to investigation of delayed fracture of welded joints on high-strength steels. Based on the results of these works, current formulation of the process of delayed fracture of welded joints on high-strength steels can be interpreted in the following way.

The nucleation of microcracks in the structure of as-quenched metal takes place locally in areas of accumulation of dislocations with high level of microstresses. As a rule, these are the areas along the grain boundaries, where metallic bonds are reduced by presence of different inclusions. The hydrogen dissolved in metal in the process of diffusion penetrates in these zones and reduces the level of surface energy necessary for microcrack nucleation. There are number of such zones and microcracks are nucleated simultaneous in several places. When the microcracks have nucleated, local relaxation of microstresses takes place. Microplastic strains on dislocation mechanism start to develop in the metal under effect of welding stresses formed in the joint during thermal-deformation cycle of welding. Movement of dislocations starts and hydrogen, traveling with them, accumulates again in the areas of their increased density blocking further slippage. Level of local microstresses in these areas of the structure rapidly increases, breaking of metallic bonds takes place, and microcrack propagates further. The microcracks, achieving their critical size, and in the presence of high level of stresses, can further propagate along a grain body and form a general microcrack in the welded joint metal. As a rule, such cracks are located in the field of main tensile stresses of the welded joint and their development is stopped on a boundary of compressive stresses.

Area of HAZ metal overheating [14–18] is the most dangerous place of the joint, where the nucleation of cold cracks has the highest possibility in arc welding of high-strength steel, the content of carbon in which exceeds 50 %. Significant changes of structure takes place in the metal due to the peculiarities of influence of thermal-deformation cycle of arc welding (surfacing) on a process of austenite homogenization and further $\gamma$-$\alpha$ transformation at cooling. As a result, a structure with increased dislocation density and high level of internal stresses is formed in the area of HAZ metal overheating. It shows that, even if content of diffusible hydrogen in the deposited metal is relatively low (to 2.2 $cm^3/100$ g), the critical stresses at delayed fracture make only 0.10–0.25 of HAZ metal yield strength in welding of high-strength steels for wheel production with carbon content 0.58 % (steel of grade 2, steel KS2). Improvement of the delayed fracture resistance requires providing of the conditions, at which not more than 50 % of martensite is formed in HAZ metal structure and limiting the content of diffusible hydrogen in the deposited metal at the level not more than 0.3 ml/100 g [15, 17].

According to GOST 10791, content of carbon in grade 2 steel makes 0.55–0.65 %. In present time, question about rise of carbon content up to 0.75 % for increase of service properties of railway wheels is considered in Ukraine and CIS countries. Similar railway wheels are manufactured in USA (wheel type C, AARM 107), Japan (JIS E 5402) and China, and EU countries (R3, ISO 1005/6). At the same time, there are no experimental data on effect of carbon content on the delayed fracture resistance of HAZ metal of grade 2 steel.

The aim of present work lied in a performance of comparative investigations on determination of effect of rise of carbon content in KS2 steel (up to 0.75 %) on the delayed fracture resistance of HAZ metal in electric arc surfacing. High-strength carbon steels of the composition (wt.%) given below were used as the material for investigations:

- grade 2 steel (GOST 10791) – 0.58 C; 0.44 Si; 0.77 Mn; 0.10 Ni; 0.05 Cr; 0.012 S; 0.011 P;
- carbon structural steel 65G (GOST 1050) – 0.065 C; 0.19 Si; 0.91 Mn; 0.18 Ni; 0.16 Cr; 0.017 S; 0.010 P;
- grade M76 steel (GOST 24182) – 0.74 C; 0.30 Si; 0.80 Mn; 0.10 Ni; 0.15 Cr; 0.012 S; 0.011 P.

**Methods of investigation.** Quantitative evaluation of the delayed fracture resistance of HAZ metal was carried out using a well-known Implant method [6]. Conditions for preparation of specimens on high-strength carbon steel and performance of the experiments were similar to those in work [17]. The maximum loading stresses of the specimens $\sigma_{cr}$, at which no delayed fracture was observed during 24 h, were taken as an evaluation criterion.

Mechanized surfacing in $CO_2$ using Sv-08G2S wire of 1.2 mm diameter was used in performance of the comparative investigations. Surfacing of the specimens was carried out using the following modes: $I_w$ = 160–180 A, $U_a$ = 26–28 V, $v_w$ = 8.1–25 m/h. At that, heat input of surfacing was changed in $Q_w$ = 4.8–15 kJ/cm range. Welding current was increased up to 220–240 A (at $v_w$ = = 8.8 m/h) for receiving of higher value of heat input (up to 21 J/cm). Surfacing was carried out with preheating ($T_{pr} \leq 250$ °C). HAZ metal cooling rate $w$ of 600–500 °C temperature interval changed in $w_{6/5}$ = 3–37 °C/s range, and 800–100 °C cooling time $\tau$ made $\tau_{8/1}$ = 110–1050 s at given conditions of surfacing. Table 1 provides for the main characteristics of thermal cycle of HAZ metal in surfacing of Implant specimens. Content of diffusible hydrogen $[H]_{dif}$ in the deposited metal, determined by method of «pencil» test using mixture of glycerin and distilled water as a locking fluid, made 0.75–0.90 ml/100 g.

Ruptures of the Implant specimens was investigated by scanning electron microscopy methods*. Philips scanning electron microscope SEM-

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

**Table 1.** Parameters of thermal cycle in HAZ metal of Implant specimens during surfacing ($T_{max}$ = 1250–1350 °C)

| $Q_w$, kJ/cm | $T_{pr}$, °C | $w_{6/5}$, °C/s | $\tau_{8/5}$, s | $\tau_{8/1}$, s |
|---|---|---|---|---|
| 4.8 | 20 | 32–37 | 6 | 110 |
| | 50 | 27–32 | 7 | 150 |
| | 70 | 25–30 | 8 | 170 |
| | 100 | 20–25 | 10 | 300 |
| | 150 | 16–18 | 12 | 650 |
| | 200 | 8–10 | 16 | 900 |
| 8.6 | 20 | 25–30 | 8 | 170 |
| | 50 | 20–25 | 10 | 230 |
| | 70 | 15–20 | 11 | 250 |
| | 100 | 12–15 | 12 | 450 |
| | 150 | 8–10 | 14 | 760 |
| | 200 | 5–7 | 18 | 890 |
| | 250 | 3–4 | 25 | 1050 |
| 11.5 | 20 | 15–17 | 14 | 210 |
| | 50 | 12–14 | 16 | 360 |
| | 100 | 6–8 | 20 | 850 |
| 15.0 | 20 | 10–12 | 17 | 290 |
| 21.0 | 20 | 7–9 | 22 | 940 |

515, equipped with energy-dispersion spectrometer of LINK system, was used.

**Results and discussion.** In the beginning the results of investigation of the delayed fracture resistance of HAZ metal of grade 2 steel will be considered in more details in order to explain the properties of HAZ metal on high-strength steels

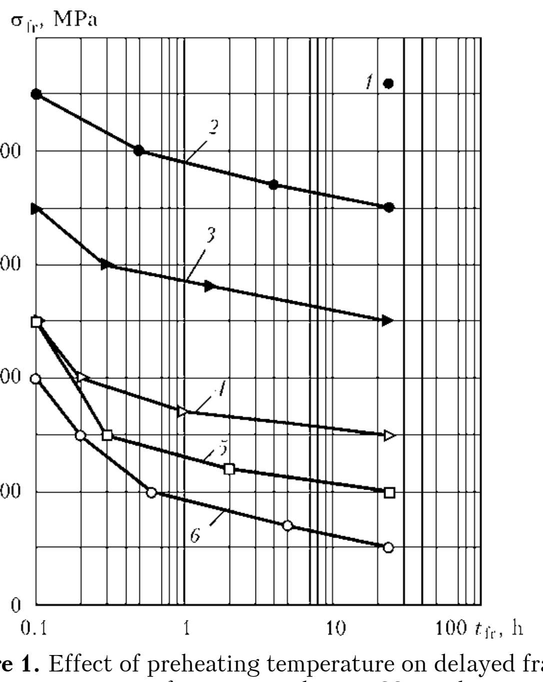


**Figure 1.** Effect of preheating temperature on delayed fracture resistance $\sigma_{fr}$ of HAZ metal on KS2 steel at $Q_w$ = = 4.8 kJ/cm: *1* — $T_{pr}$ = 200; *2* — 150; *3* — 100; *4* — 70; *5* — 50; *6* — 20 °C

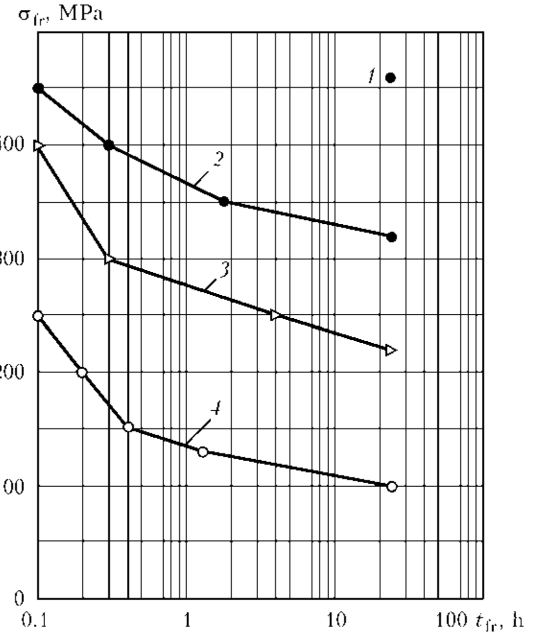


**Figure 2.** Effect of preheating temperature on delayed fracture resistance $\sigma_{fr}$ of HAZ metal on KS2 steel at $Q_w$ = = 8.6 kJ/cm: *1* — $T_{pr}$ = 100; *2* — 70; *3* — 50; *4* — 20 °C

with higher carbon content (steels 65G and M76). Figures 1 and 2 show the effect of preheating temperature on indices of critical stresses during testing of Implant specimens, the surfacing of which was carried out at 4.8 and 8.6 kJ/cm heat input, respectively. Figure 3 shows a diagram of transformation of undercooled austenite in HAZ metal of KS2 steel in arc surfacing [15].

As can be seen from given material, structural condition of the metal in HAZ overheating area has significant effect on change of indices of the delayed fracture resistance $\sigma_{cr}$. Martensite-bainite structure, the content of martensite in which exceeds 71 %, is formed in the metal of HAZ overheating area at cooling rate $w_{6/5}$ 25 ≥ °C/s.

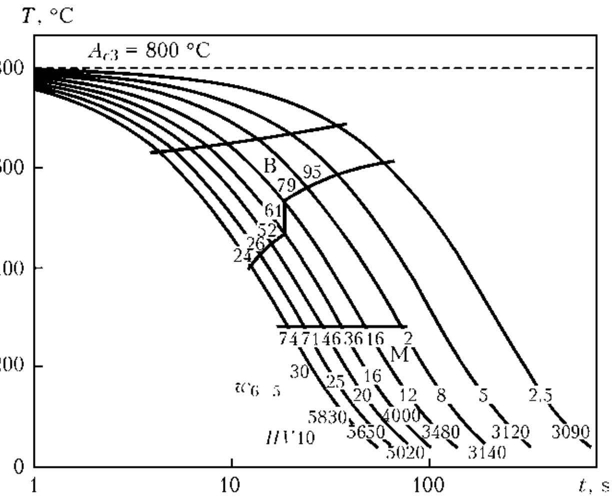


**Figure 3.** Diagram of transformation of undercooled austenite in HAZ metal on steel KS2 (0.58 % C) in arc surfacing at $w_{heat}$ = 210 °C/s and austenitizing time $t_1$ = = 7–10 s [15]

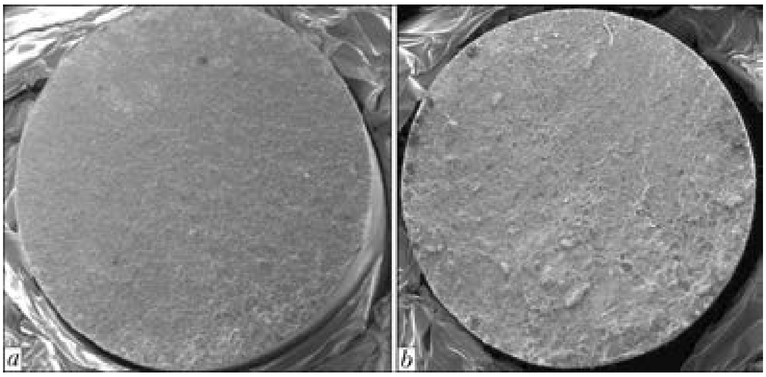

**Figure 4.** Ruptures (×24) of Implant specimens of KS2 steel in surfacing at 8.6 kJ/cm heat input without (*a*) and with (*b*) preheating to 70 °C

Martensite microhardness rises from $HV0.1$ = = 4420 MPa at $w_{6/5}$ 8 °C/s up to 5660–6060 MPa at 25–30 °C/s. Integral hardness of metal makes $HV10 \geq 5650$ MPa. Therefore, HAZ metal has low resistance to nucleation and propagation of the microcracks, and critical fracture stresses make only $0.07\sigma_{0.2}$ and $0.14\sigma_{0.2}$, respectively, in surfacing without preheating at $Q_w$ = = 4.8 and 8.6 kJ/cm. In order to eliminate the development of delayed fracture process in the metal of HAZ overheating area under given conditions of surfacing, the external stresses should not exceed 50–100 MPa.

Using of preheating at $T_{pr}$ = 50 °C provides significant reduction of cooling rate to $w_{6/5} \geq$ ≥ 20 °C/s (see Table 1), and martensite-bainite structure with lower volume fraction of martensite (50–74 %) is also formed in HAZ metal. Critical stresses rise virtually 2 times, but still remain at the low level. Index $\sigma_{cr} = 0.14\sigma_{0.2}$ in surfacing at 4.8 kJ/cm heat input and it makes $\sigma_{cr} = 0.30\sigma_{0.2}$ at 8.6 kJ/cm. At that realization of the process of delayed fracture of HAZ metal (time of fracture $t_{fr}$ = 0.1 h) requires the external stresses of 250 and 400 MPa value, respectively.

Application of preheating at 70 °C and higher promotes more significant increase of delayed fracture resistance of deposited HAZ metal, received at 8.6 kJ/cm, up to $\sigma_{cr} \geq 0.45\sigma_{0.2}$. The structure of HAZ metal under given conditions of surfacing ($T_{pr}$ = 70–100 °C) and depending on $w_{6/5}$ = 12–20 °C/s (see Table 1) is represented by bainite-martensite mixture, the volume fraction of martensite in which does not exceed 50 %. When using surfacing at 4.8 kJ/cm, such conditions of structure formation correspond to application of preheating at 150–200 °C, and critical stresses in fracture of HAZ metal increase up to $\sigma_{cr} \geq 0.50\sigma_{0.2}$. Testing of Implant specimens at higher load values failed due to beginning of a metal plastic flow and impossibility of realization of the delayed fracture process. Therefore, the critical stresses were conditionally taken equal 460 MPa in surfacing with $Q_w$ = 8.6 kJ/cm and $T_{pr}$ = 100 °C, as well as $Q_w$ = 4.8 kJ/cm and $T_{pr}$ = 200 °C.

Fracture of HAZ metal of KS2 steel at all investigated variants of surfacing, described above, has brittle nature. Figure 4 shows ruptures of the Implant specimens, surfacing of which was carried out at 8.6 kJ/cm heat input with or without preheating to 70 °C, and Figure 5 displays typical fragments of fracture surface in areas of nucleation and propagation of microcracks. Loading at specimen fracture made $\sigma_{fr} = (1.1–1.3)\sigma_{cr}$.

Typical areas of fracture are observed on the rupture surfaces. They can be symbolically divided on local areas of nucleation of microcracks and their delayed propagation (zone I), areas of rapid propagation of microcrack to macrolevel (zone II) and areas of final rupture (zone III). Distribution of zones has local nature, nucleation and propagation of cracks take place simultaneously in several places, areas of fracture alternate. Microlevel investigations showed that nucleation of microcrack in HAZ for both variants of surfacing of KS2 steel has brittle nature and takes place along the grain boundaries (Figure 5, *a*, *b*). Further, the microcracks propagate along the boundaries as well as in grain body. The rupture in this zone is characterized as intergranular and transcrystalline cleavage. Size of cleavage facets is $D_f$ ~ 30–100 μm.

Relationship of these types of rupture will change depending on HAZ metal structural condition. When surfacing without preheating, the volume fraction of intergranular cleavage makes 95 %, and with preheating at 70 °C it reduces to 30 %. Areas of quasi-brittle fracture and local areas of tough constituent with $D_f$ ~ 0.5–2.0 μm in amount up to 10 % (Figure 5, *c*, *d*) are also observed along the grain boundaries in preheating application.

Nature of the final rupture and size of structural elements differ from other zones of fracture.

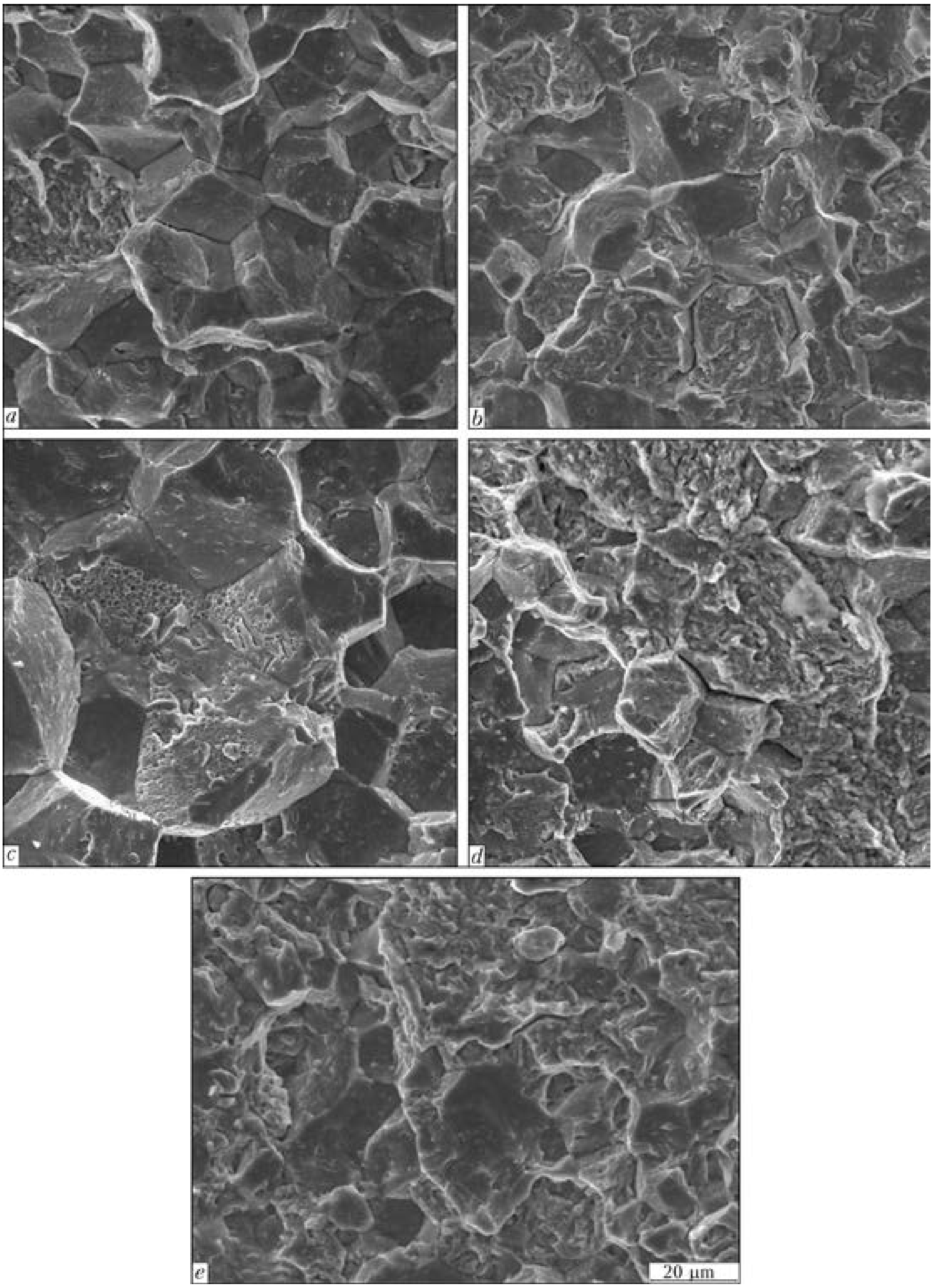


**Figure 5.** Fragments (×1010) of fracture surface in area of nucleation (*a*, *b*), propagation of microcrack (*c*, *d*) and in zone of final rupture (*e*) of HAZ metal on steel KS2 in surfacing with $Q_w$ = 8.6 kJ/cm: *a*, *c* — $T_{pr}$ = 20; *b*, *d* — 70 °C

It is similar for the both variants of surfacing. It is mainly intergranular quasi-brittle fracture with $D_f$ ~ 10–20 μm (Figure 5, *e*).

Increase of heat input of surfacing more than 8.6 kJ/cm provides rise of resistance to the delayed fracture of HAZ metal on KS2 steel specimens (Figure 6). Thus, cooling rate reduces to 15–17 °C/s in surfacing with $Q_w$ = 11.5 kJ/cm without preheating. The same bainite-martensite structure as in surfacing with $Q_w$ = 11.5 kJ/cm and $T_{pr}$ = 70 °C is formed in the HAZ metal, and critical stresses make 200 MPa. But this $\sigma_{cr}$ value is approximately 1.6 times lower than the similar index for surfacing with lower heat input at approximately the same rate of cooling in HAZ. Apparently, it is related with the development of process of growth of austenite grain in the metal of HAZ overheating area, which is more intensive in surfacing at higher heat input. However, even if this negative effect is taken into account, application of preheating at 50 °C in surfacing with $Q_w$ = 11.5 kJ/cm eliminates propagation of the delayed fracture in HAZ metal on KS2 steel. It should also be noted that testing of Implant specimens in surfacing at higher heat input showed increase of time before fracture. Thus, delayed fracture of the specimens at $\sigma_{fr} = 2\sigma_{cr}$, surfaced with $Q_w$ = 8.6 kJ/cm and $T_{pr}$ = = 70 °C, continued 0.1 h, and fracture period at $Q_w$ = 11.5 kJ/cm and $T_{pr}$ = 20 °C made 0.7 h. This also can be related with growth of grain in HAZ metal and corresponding enlargement of path of propagation of microcrack along its boundary, location of which does not match with the field of action of main tensile stresses.

No development of the process of delayed fracture was registered in surfacing without preheating of KS2 steel (0.58 % C) at 15 kJ/cm heat input. In this case, the rate of cooling in HAZ makes 10–12 °C/s, and bainite-martensite structure with volume fraction of martensite not exceeding 16 % is formed. Hardness of metal in HAZ overheating areas makes *HV*10 ≤ 3480 MPa. The microcracks are not nucleated in such a metal due to development of the processes of microplastic strain and relaxation of microstresses in the structure.

The delayed fracture of HAZ metal takes place at lower values of external stresses with the increase of carbon content in high-strength steel. Figure 7 represents the generalized results of Implant specimen testing. As can be seen from data presented, the increase of heat input up to 15 kJ/cm and above in surfacing without preheating of 65G (0.65 % C) and M76 (0.74 % C) steels promotes rise of critical stresses, but does not eliminate development of the process of delayed fracture in HAZ metal.

Specially carried out metallographic investigations* showed that the structure of 65G structural steel in as-delivered condition is represented by bainite (Figure 8, *a*), grain size is 16–24 μm and integral hardness of metal makes *HV*10 = = 2760 MPa. The bainite-martensite structure with volume fraction of martensite not exceeding 50 % is formed in the metal of HAZ overheating area on steel 65G under effect of TCW at cooling rate $w_{6/5}$ ≤ 7 °C/s, in contrast to steel KS2. Hardness of the metal with such a structure is not more than 4570 MPa. Size of grain in the metal of HAZ overheating area makes 63–94 μm. Increase of cooling rate up to 12–15 °C/ s promotes formation in HAZ metal of a structure with

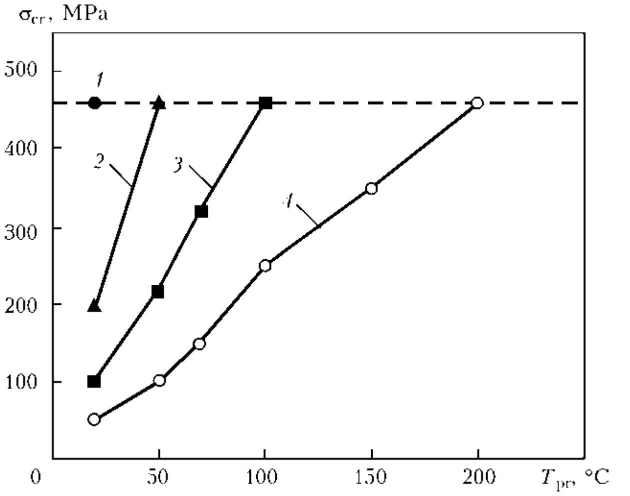


**Figure 6.** Effect of heat input of surfacing and preheating temperature on delayed fracture resistance of HAZ metal on steel KS2 (0.58 % C): *1* — $Q_w$ = 15.0; *2* — 11.5; *3* — 8.6; *4* — 4.8 kJ/cm

volume fraction of martensite making more than 97 %. Lower bainite (2 %) is also present in structure of the metal at given cooling rate. It is locally situated along the grain boundaries. Microhardness of martensite depending on cooling rate makes *HV*0.1 = 4250–7390 MPa, and integral

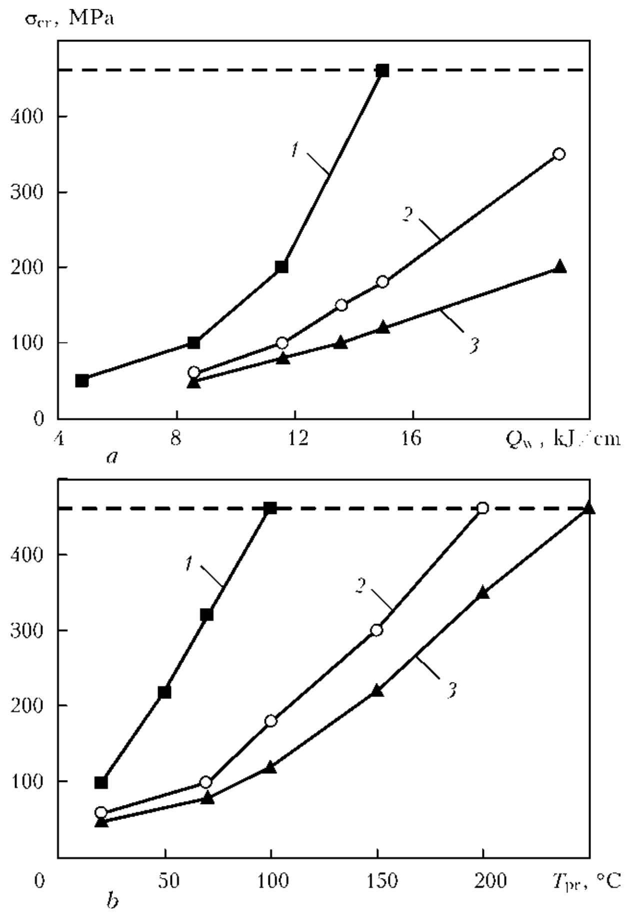


**Figure 7.** Effect of heat input of surfacing (8.6 kJ/cm) (*a*) and temperature of preheating (20 °C) (*b*) on resistance to delayed fracture of high-strength carbon steel: *1* — steel KS2 (0.58 % C); *2* — 65G (0.65 % C); *3* — M76 (0.74 % C)

*Investigations were carried out by V.A. Kostin, V.V. Zhukov and T.G. Solomijchuk.

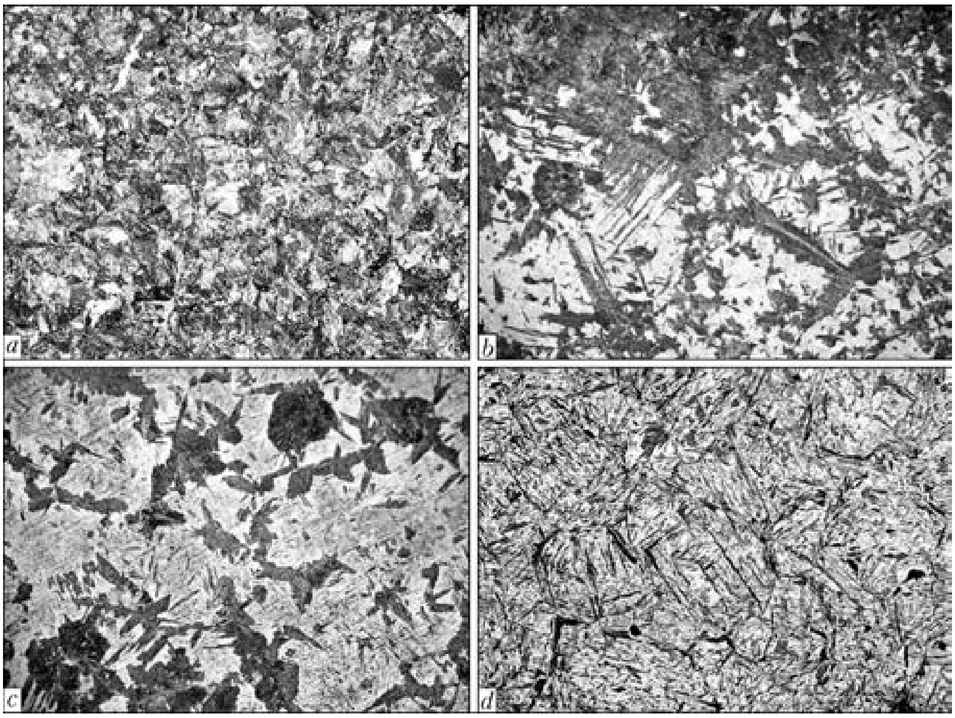

**Figure 8.** Microstructure of area of HAZ metal overheating on steel 65G (×500, reduced 2 times): *a* — as-delivered steel (×200); *b* — $w_{6/5}$ = 5; *c* — 8; *d* — 20 °C/s

hardness of the metal is to 7200 MPa (Figure 8, *d*). Microhardness of martensite makes 5600–6130 MPa at $w_{6/5}$ = 8 °C/s.

Structure of HAZ metal of 65G steel respectively influences the change of its resistance to delayed fracture. The critical stresses make more than 350 MPa in formation in HAZ metal of relatively plastic bainite-martensite structure. The latter is formed at $w_{6/5} \leq 7$ °C/s that corresponds to application of heat input of surfacing on the level of $Q_w \geq 21$ kJ/cm or $T_{pr}$ >150 °C at $Q_w$ = 8.6 kJ/cm. At that, the increased resistance of HAZ metal to delayed fracture for steel 65G as well as for KS2 steel is provided at $\sigma_{cr} \geq$ $\geq 0.45\sigma_{0.2}$ during formation of the bainite-martensite structure, the volume fraction of martensite in which does not exceed 50 %. If structure with high content of martensite is formed in HAZ metal, its resistance to the delayed fracture reduces more than 3 times.

Study of ruptures of HAZ metal of 65G steel specimens, the generalized results of which are given in Figures 9 and 10 and Table 2, showed that the increase of heat input of surfacing from 13.6 to 21 kJ/cm, when cooling rate in HAZ metal reduces approximately from 15 to 7 °C/s, promotes significant changes in fracture nature. Similar to surfacing of KS2 steel, the microcracks in HAZ metal are nucleated and then slowly propagate along the boundaries as well as in grain body. The rupture in this zone is characterized as intergranular and transcrystalline cleavage (see Figure 10). Volume fraction of the intergranular cleavage, which makes 85 % at 13.6 kJ/cm, reduces to 50 % at 21 kJ/cm with increase of the surfacing heat input. At that, size of facets on the surface of fracture increases from 25–50 to 50–100 μm. Fraction of brittle intergranular cleavage is also reduced in the zone of accelerated propagation of microcrack (zone II), and local areas of tough constituent (up to 10 %) are observed along the grain boundaries.

If criterion of increased resistance to the delayed fracture of HAZ metal for studied KS2 and

**Table 2.** Structure of rupture during delayed fracture of HAZ metal on 65G steel

| $Q_w$, kJ/cm | Zone I | | | | Zone II | | | | | |
|---|---|---|---|---|---|---|---|---|---|---|
| | Brittle intergranular | | Brittle transcrystalline | | Brittle intergranular | | Brittle transcrystalline | | Tough | |
| | *V*, % | $D_f$, μm | *V*, % | $D_f$, μm | *V*, % | $D_f$, μm | *V*, % | $D_f$, μm | *V*, % | $D_f$, μm |
| 13.6 | 85 | 25/50 | 15 | 25/50 | 60 | 30/50 | 40 | 30/70 | – | – |
| 15.0 | 70 | 30/70 | 30 | 30/70 | 40 | 30/70 | 60 | 40/70 | – | – |
| 21.0 | 50 | 50/100 | 50 | 50/100 | 20 | 30/100 | 70 | 40/70 | 10 | 0.5/2.0 |

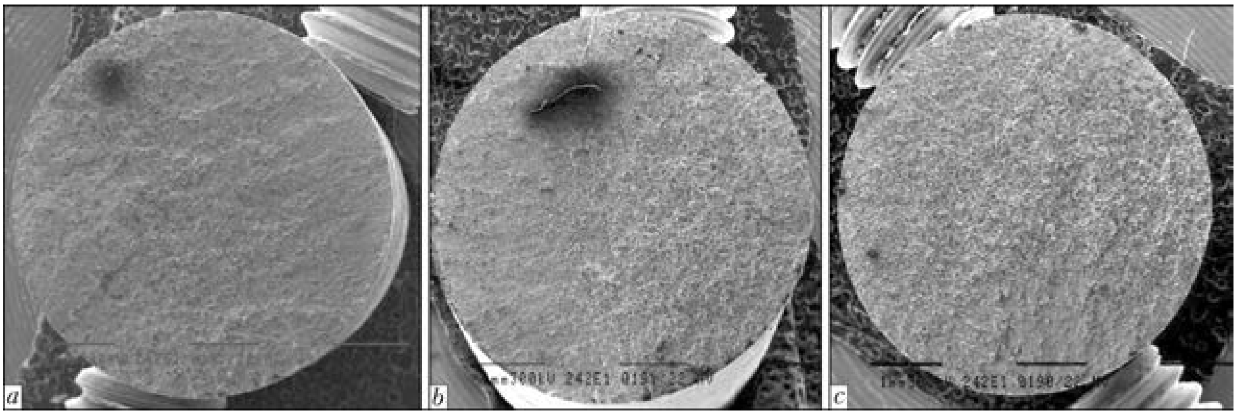

**Figure 9.** Ruptures (×15) of Implant specimens on steel 65G in surfacing without preheating: *a* — $Q_w$ = 13.6; *b* — 15.0; *c* — 21.0 kJ/cm

65G steels, carbon content in which does not exceed 0.65 %, was provided in formation of the bainite-martensite structure, the volume fraction of martensite in which does not exceed 50 %, then properties of M76 steel (0.74 % C) differ from previous ones. Temperature of beginning of austenite to martensite transformation ($T_{b.M}$) in HAZ metal of steel M76 under effect of arc surfacing thermal cycle makes 240–250 °C, depending on development of transformations in transitional region, and it is more than 40 °C lower of $T_{b.M}$ temperature for KS2 (0.58 % C) and 65G (0.65 % C) steels. Microhardness of the structure martensite constituent, depending on cooling rate, is at the level of $HV0.1$ = 5030–7880 MPa. Obviously, that the local microstresses in martensite structure of M76 steel will be significantly higher than in steels with lower content of carbon. Therefore, nucleation and propagation of microcracks should take place at lower volume fraction of martensite in the structure and at lower values of external stresses. It is verified by the results of testing of Implant specimens (see Figure 7). As can be seen, the increaed delayed

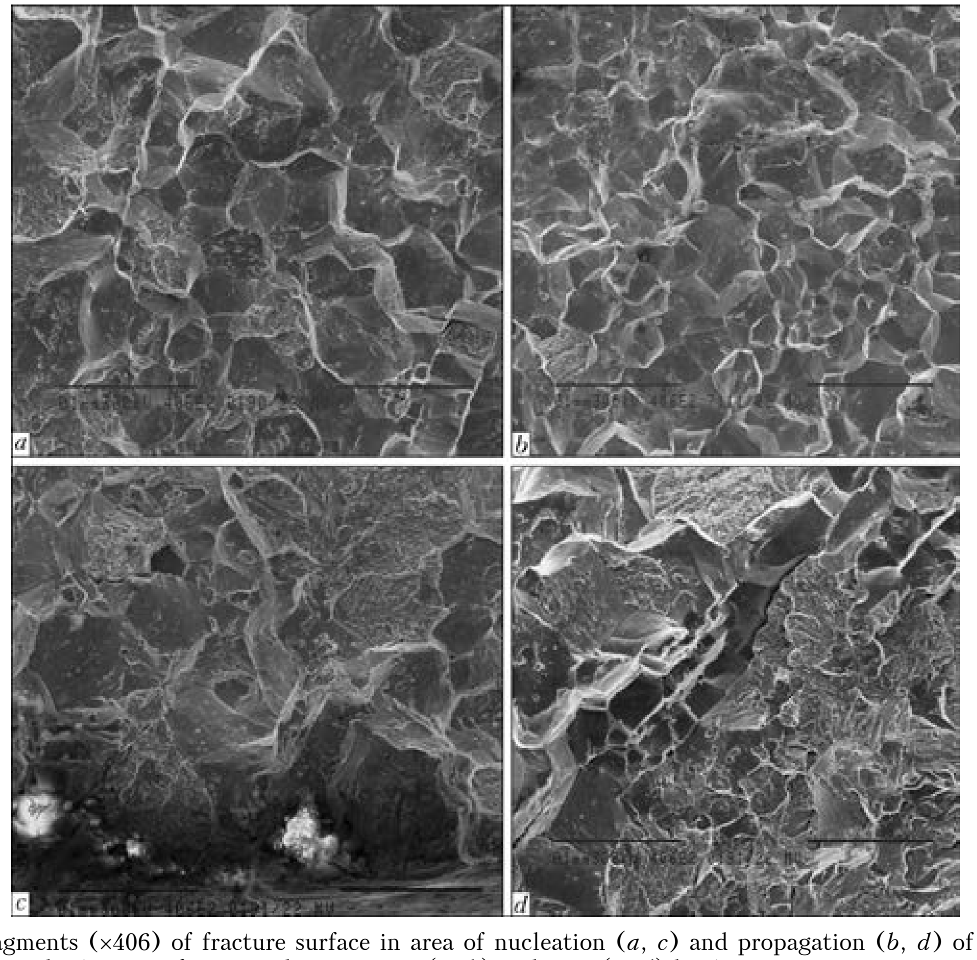

**Figure 10.** Fragments (×406) of fracture surface in area of nucleation (*a*, *c*) and propagation (*b*, *d*) of microcrack in HAZ metal on steel 65G in surfacing with $Q_w$ = 13.6 (*a*, *b*) and 21.0 (*c*, *d*) kJ/cm

fracture resistance of HAZ metal on steel M76 at $\sigma_{cr} \geq 0.45\sigma_{0.2}$ level can be provided using preheating temperature above 200 °C, when the rate of cooling makes not more than 5–7 °C/s. It was determined that the bainite-martensite structure is formed at this cooling rate, volume fraction of martensite in it does not exceed 10 % and its microhardness $HV0.1$ = 5030–6200 MPa. Lower bainite with 3360–3780 MPa microhardness is the prevailing structure at $w_{6/5} \leq 12$ °C/s. No delayed fracture of HAZ metal on M76 steel takes place at $T_{pr}$ = 250 °C, when $w_{6/5}$ = 3– 4 °C/s and content of martensite in the structure is not more than 2 %.

Generalizing the results of investigations given above, the effect of carbon content in KS2 steel on HAZ metal resistance to delayed fracture can be represented in form of diagram $w_{6/5} = f(C)$. Three typical areas of HAZ metal susceptibility to delayed fracture (Figure 11) are outlined:

1. Area of active development of the process of delayed fracture in HAZ metal. If the cooling rate $w_{6/5}$, indicated by upper curve in the diagram, is exceeded, then the hardening structures, susceptible to nucleation and propagation of the cold cracks at presence of sufficiently low external stresses ($\sigma_{cr} < 0.30\sigma_{0.2}$), will be formed in HAZ metal. The structures, in which the volume fraction of martensite is more than 50 %, are formed at carbon content in wheel steel in the limits of 0.55–0.65 %, and structure with 10 % of martensite and more is developed at 0.65–0.75 % C. In this area the process of nucleation of microcracks and their development to macrolevel is realized during 0.1 h, and its elimination is virtually impossible.

2. Area of increased delayed fracture resistance of HAZ metal. Nucleation and propagation of microcracks are complicated due to formation of relatively plastic structures. This is HAZ metal structure, in which the volume fraction of martensite makes more than 28–36 %, but does not exceed 50 % at carbon content in KS2 steel up to 0.65 %. The structure includes from 4 to 10 % of martensite at 0.65–0.75 % C. The nucleation of microcracks is possible only under condition of presence of stresses at the level of $\sigma_{cr} \geq 45\sigma_{0.2}$ with continuous cooling of metal in low-temperature area ($T \leq T_{b.M}$), when the processes of microplastic strain take place in completely formed as-quenched structure. This area of cooling rates is limited by two curves in the diagram. The process of delayed fracture in this case is realized during longer period of time (more than 0.7 h). Delayed fracture of the deposit HAZ metal can be prevented by slowing its cooling in the area of martensite transformation temperatures ($T \leq T_{b.M}$). Practically, it can be performed using controlled thermal cycle of surfacing.

3. Area, in which the process of delayed fracture of KS2 steel in electric arc surfacing is not realized. When providing the cooling rate, which does not exceed the lower boundary of indicated area of increased resistance, the structures, having sufficiently high susceptibility to microplastic strain without microcrack nucleation, are formed in the HAZ. The cooling rate is $w_{6/5} \leq 10$ °C/s at content of carbon in KS2 steel at the level of 0.60 %, $w_{6/5} \leq 5$ °C/s at C ≤ 0.65 % and $w_{6/5} \leq 3$ °C/s at C ≤ 0.75 %.

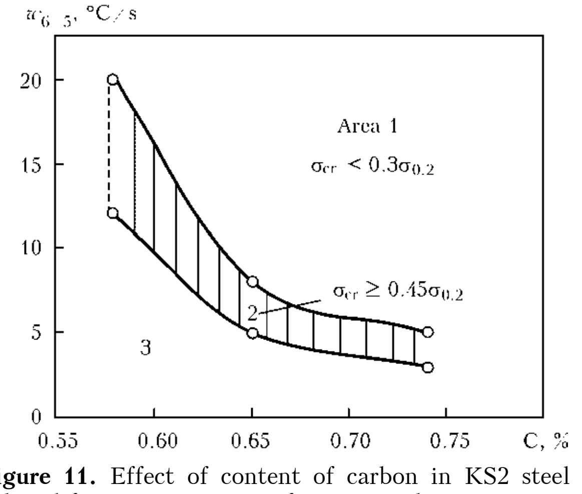


**Figure 11.** Effect of content of carbon in KS2 steel on delayed fracture resistance of HAZ metal

It should also be noted that the diagram given in Figure 11 meets the conditions of electric arc surfacing, when content of diffusible hydrogen in the deposited metal does not exceed 0.75–0.90 ml/100 g. When its content increases, that depends on methods and parameters of surfacing, system of alloying of surfacing consumables and methods of their preparation etc., the special technological measures on reduction of its negative effect [17] should be applied.

Practical recommendations on selection of temperature of preheating and heat input of surfacing are realized in technologies for retailoring of worn flanges of solid-rolled wheels of freight cars. Today, the repair enterprises of railway transport of Ukraine during repair of wheels, made from KS2 steel, use a technology of restoration applying submerged dual-arc surfacing developed at the E.O. Paton Electric Welding Institute of the NASU in cooperation with Design Bureau «Ukrzaliznytsya» Company. Surfacing is carried out using Sv-08KhM wire at heat input 11–13 kJ/cm. Mandatory elements of the technology are application of preheating of wheel treads to 150 °C and delayed cooling of the wheels after surfacing in thermal chambers, where they are soaked during 5 h (cooling rate makes around 50 °C/h). High quality of deposited metal and safety of wheels in operation is achieved with fulfilment of set of requirements, made to the surfacing technologies, in particular, to process equipment, consumables, procedure and modes of surfacing and maintaining of thermal cycle during restoration.

**Conclusions**

1. High-strength steel, content of carbon in which makes 0.55–0.75 %, have increased susceptibility to the delayed fracture of HAZ metal in electric arc surfacing due to formation of hardening structures with increased dislocation density and high level of local internal stresses. Structural condition of the metal, which depends on carbon content and conditions of cooling in surfacing, makes significant effect on the processes of nucleation and propagation of microcracks in the metal of HAZ overheating area. The increase of carbon content in steel and cooling rate in 600–500 °C temperature interval provokes rise of susceptibility to the delayed fracture of HAZ metal, and level of critical stresses reduces to $0.07\sigma_{0.2}$.

2. Formation of more plastic structures with volume fraction of martensite not exceeding 50 % promotes increase of the delayed fracture resistance of HAZ metal on grade 2 steel at carbon content 0.55–0.65 % up to $\sigma_{cr} \geq 0.45\sigma_{0.2}$ level of stresses. The volume fraction of martensite in HAZ metal structure should make not more than 4–10 % at 0.75 % C. Nucleation and propagation of microcracks in surfacing under such conditions of structure formation can be prevented by slowing down the cooling of HAZ metal in area of temperatures of martensite transformation ($T \leq T_{\text{b.M}}$).

3. Diagram in form of $w_{6/5} = f(\text{C})$ was plotted for showing the influence of carbon content in KS2 steel on HAZ metal resistance to the delayed fracture. It was determined that the structures, having sufficiently high capability to microplastic strain without generation of microcracks, are formed in the metal of HAZ overheating area for KS2 steel, content of carbon in which is 0.55–0.60 % at cooling rate $w_{6/5} \leq 16$ °C/s ($w_{6/5} \leq 8$ °C/s at 0.60–0.65 % C and $w_{6/5} \leq 5$ °C/s at 0.65–0.75 % C.


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