INFLUENCE OF THERMODEFORMATIONAL CYCLE OF HARDFACING ON THE STRUCTURE AND PROPERTIES OF RAILWAY WHEELS AT THEIR RECONDITIONING

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Influence of preheating temperature on delayed fracture of HAZ metal has been assessed. It is established that in order to ensure a high resistance to delayed fracture, preventing cold cracking in the joint HAZ metal, the cooling rate $w_{6/5}$ should not be higher than 5 °C/s. In this case it is necessary to apply preheating at the temperature above 200 °C.

Keywords: *arc hardfacing, solid-rolled carriage wheels, HAZ, structure, mechanical properties*

Solid-rolled wheels for railway transportation are made of carbon steels subjected to special heat treatment to achieve the required strength and wear resistance. Comparative composition and mechanical properties of wheel steels are given in Tables 1 and 2. Carbon content in steels is equal to 0.44–0.67 wt.%, and strength level is higher than 900 MPa. Wheels wear along the rolling profile in service. In view of the features of rolling-friction wheel–rail pair, mostly the wheel flange working surface is exposed to wear. The railway transportation enterprises apply the technologies of flange hardfacing to restore the wheel rolling profile, which is cost-effective. Repairing flange wear by hardfacing allows reducing the waste of rim metal at its mechanical turning along the rolling profile, as well as improving the wheel wear resistance through deposition of metal with preset properties

To restore solid-rolled wheels of freight transportation, made from steel 2, the technologies of single- and twin-submerged-arc welding with solid wires are used [1, 2]. These technologies are based on the results of investigation of weldability of high-strength carbon steels [2–4]. To recondition the flanges of solid-rolled wheels technological recommendations, as well as special hardfacing and auxiliary equipment have been developed. Submerged-arc hardfacing of wheel flanges is performed in the modes, providing heat input on the level of 10–14 kJ/cm. The technology envisages compulsory application of preheating of wheel rims up to the temperature of 150–200 °C (depending on the used hardfacing process) and postweld delayed cooling of wheels in heat chambers at not more than 50 °C/h rate. In case of fulfillment of a set of requirements made to the hardfacing technology, in particular, process equipment, welding consumables, technique and modes of hardfacing, maintaining the thermal cycle during reconditioning, a high quality of the deposited metal and wheel reliability in operation are guaranteed.

Before 2006, the rolling stock of railway freight transportation in Ukraine and CIS countries was fitted with solid-rolled wheels, made from steel of type 2 with carbon content closer to the lower limit (C ≤ ≤ 0.60 wt.%). Starting from 2006, in order to reduce the wear and improve the performance, solid-rolled carriage wheels began to be made from new wheel

Table 1. Composition of wheel steels, wt.%

Wheel steel type	C	Mn	Si	V	S, not more than	P, not more than	C_{eq}, wt.%
1 (GOST 10791–89)	0.44–0.52	0.80–1.20	0.40–0.60	0.08–0.15	0.030	0.035	0.68
2 (GOST 10791–89)	0.55–0.65	0.50–0.90	0.22–0.45	≤ 0.10	0.030	0.035	0.73
T (TU U 35.2-23365425-600:2006)	0.58–0.67	0.70–0.90	≤ 0.40	0.08–0.15	0.020	0.025	0.79

Table 2. Mechanical properties of wheel steels

Wheel steel type	*HB*	σ_t, MPa	δ_5, %, not less than	ψ, %, not less than	KCU_{+20}, J/cm^2
1 (GOST 10791–89)	≥ 248	900–1100	12	21	30
2 (GOST 10791–89)	≥ 255	930–1130	8	14	20
T (TU U 35.2-23365425-600:2006)	≥ 320	≥ 1100	8	14	18

steel of T type, the ultimate strength of which is $\sigma_t \geq$ $\geq$ 1100 MPa, and hardness is more than *HB* 320. Unlike wheel steel of type 2 the new steel has a higher content of carbon (up to 0.67 wt.%) and vanadium (up to 0.15 wt.%). Carbon equivalent C_{eq} of wheel steel of type T is equal to 0.79 wt.%. Therefore, welded joints of the above wheel steel have a higher cold cracking susceptibility, compared to wheel steel of type 2 ($C_{eq} \geq 0.73$ wt.%) [5].

This leads to the suggestion that the hardfacing technologies which are used with success at reconditioning of flanges of wheels from type 2 steel cannot be used for repair of higher strength wheels. Hence the need for development of a new technology of hardfacing repair of flanges of railway wheels from wheel steel of type T, which made it necessary to study its weldability.

This work gives the results of experimental studies on assessment of the influence of thermodeformational cycle of welding on the structure, mechanical properties and delayed fracture resistance of HAZ metal of joints of new wheel steel. These studies were performed on wheel steel of type T of the following composition, wt.%: 0.625 C; 0.73 Mn; 0.31 Si; 0.11 V.

Structure of HAZ metal of welded joints made by arc welding processes is heterogeneous, and dimensions of its individual sections are small. Therefore, structural changes, occurring under the impact of thermodeformational cycle of welding and their influence on mechanical properties in the HAZ overheated zone were studied on model samples. For this purpose, Gleeble 3800 system fitted with a thermostat and high-speed dylatometer was used [6]. Studied cylindrical samples had the diameter of 6 mm and length of 80 mm. In keeping with the testing procedure, they were heated up to the temperature of 1200 °C at the rate of 150 °C/s, and then cooled by different thermal cycles, which were selected so that in the temperature range of 600–500 °C the sample cooling rate $w_{6/5}$ changed in the range from 2.85 to 33 °C/s. Temperature of the start and end of overcooled austenite transformation was determined by the tangent moving away from the dylatometric curve, and the ratio of phases formed as a result of austenite transformation, was established by the line-segment approach [7]. Figure 1 gives the generalized investigation results.

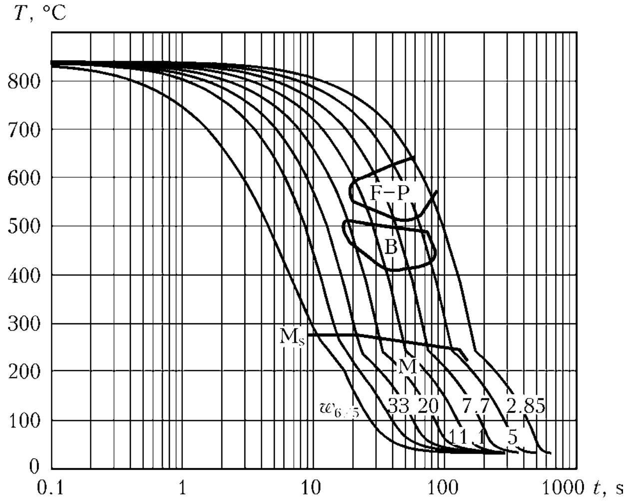

Figure 1. Thermokinetic diagram of overcooled austenite transformation in HAZ metal of wheel steel of type T (C = 0.625 wt.%): M — martensite; M_s — start of martensite transformation; B — bainite; F — ferrite; P — pearlite

As shown by the conducted studies, in wheel steel of type T under the impact of the thermodeformational cycle of welding at cooling rate $w_{6/5} \geq 33$ °C/s overcooled austenite transformation occurs solely in the martensite region. Temperature of the start of martensite transformation is equal to 280 °C, and quenched metal microhardness is *HV*0.5 594. With lowering of the cooling rate the metal structure forms intermediate phases of ferrite-pearlite mixture and bainite, with lowering of martensite content. At $w_{6/5}$ = 20 °C/s (Figure 2, *a*) an intermediate bainite transformation of overcooled austenite occurs in the temperature range of 510–450 °C, and martensite transformation starts at 270 °C. The proportion pf phases in the structure is equal to 96 vol.% of martensite and 4 vol.% of bainite. Quenched metal microhardness decreases to *HV*0.5 500.

At $w_{6/5}$ = 11.1 °C/s (Figure 2, *b*) overcooled austenite transformation starts in the ferrite-pearlite region at 500–410 °C with bainite formation, martensite transformation starts at the temperature of 260 °C. Proportion of phases in the structure is as follows, vol.%: 83 martensite, 14 bainite and 3 ferrite-pearlite mixture. Quenched metal microhardness is *HV*0.5 420. At $w_{6/5}$ = 7.7 °C/s (Figure 2, *c*) volume fraction of martensite component in the structure decreases to 60 %, with 13 % bainite, and ferrite-pearlite

Figure 2. Microstructures (×500) of HAZ metal of wheel steel of type T at $w_{6/5}$ = 20.0 (*a*), 11.1 (*b*) and 7.7 (*c*) °C/s

Table 3. Influence of cooling rate on mechanical properties of HAZ metal of wheel steel of type T (C = 0.625 wt.%)

$w_{6/5}$, °C/s	σ_t, MPa	σ_y, MPa	δ_5, %	ψ, %	*KCU*, J/cm^2	
					+20 °C	−40 °C
5	1140	850	6.3	16.3	7.3	8.5
10	1280	940	3.4	9.6	5.7	4.2
20	1320	980	3.1	9.6	6.0	4.2

Table 4. Influence of cooling rate on mechanical properties of HAZ metal of wheel steel of type 2 (C = 0.55 wt.%) [2]

$w_{8/7}$, °C/s	σ_t, MPa	σ_y, MPa	δ_5, %	ψ, %	KCU_{-40}, J/cm^2
1.15	940	600	13.3	33.3	6
5.90	970	605	12.9	33.3	6
32.0	1060	715	9.3	24.9	5

volume fraction is equal to 27 %. Quenched metal microhardness decreases to *HV*0.5 406. At $w_{6/5}$ = = 5 °C/s transformation of overcooled austenite occurs in the ferrite-pearlite region already by 77 % at 620–520 °C. At such a cooling rate, martensite content decreases abruptly, and is equal to just 10 vol.%, that of bainite is 13 vol.%, metal microhardness being equal to *HV*0.5 360. At $w_{6/5}$ = 2.85 °C/s overcooled austenite transformation occurs with formation of just the ferrite-pearlite mixture (*HV*0.5 358).

With lowering of the cooling rate martensite content in the HAZ metal structure decreases, volume fraction of ferritic-pearlitic component increases, and bainite content is stabilized on the level of 13–14 % with microhardness decreasing more than 1.5 times.

Assessment of the influence of cooling rate on the mechanical properties and impact toughness of wheel steel of type T was performed using model samples treated by the thermodeformational cycle of welding in MSR-75 unit [8]. For this purpose base metal samples of 120 × 12 × 12 mm size were heated by passing current up to 1200 °C at the rate of 150 °C/s. Sample cooling rate was 5, 10 and 20 °C/s. Then the heat-treated blanks were used to cut out standard samples for static tension testing of type II to GOST 1497–84 and impact bending of type I to GOST 9454–78. Table 3 gives mechanical properties of simulated HAZ metal of higher-strength wheel steel. For comparison Table 4 gives the data on the influence of cooling rate on the properties of wheel steel of type 2 with carbon content of 0.55 wt.% [2].

As a result of investigations it was found that at $w_{6/5} \geq 10$ °C/s, when martensite is the main structural component, HAZ metal of joints of wheel steel of T type features increased strength and low ductility. Compared to steel of type 2 strength of quenched metal is 25–30 % higher, and ductility is almost 3 times lower. In addition, lowering of the ductility level at increase of cooling rate and formation of martensite structure of the metal also runs more intensively. While the values of relative elongation δ_5 and reduction ψ of HAZ metal of wheel steel of type 2 decrease by 30 % at increase of cooling rate, in steel of type T quenched metal ductility decreases by 45 %. Here even at relatively small (5 °C/s) cooling rate ductility values of HAZ metal of wheel steel of type T are 1.5 times lower than those of wheel steel of type 2 at high (32 °C/s) cooling rate.

Influence of thermodeformational cycle of welding on the change of impact toughness of new wheel steel is manifested in a similar fashion. It is established that at increase of cooling rate up to 5–20 °C/s impact toughness of HAZ metal at negative temperature decreases almost 2 times. It should be noted that for wheel steel of type 2 at increase of cooling rate to 32 °C/s impact toughness values are lowered by 17 %.

At hardfacing of flanges of solid-rolled wheels tensile residual stresses are formed in the HAZ metal, their maximum level in the longitudinal direction reaching 650 MPa [2]. Therefore, in order to ensure the high cold cracking resistance, it is first of all necessary for wheel steel HAZ metal to have a sufficient ductility margin. This promotes a more complete running of the processes of relaxation of local stresses due to development of microplastic deformations, thus essentially increasing the HAZ metal resistance to delayed cracking [9, 10]. Conducted studies showed that the ductile properties of HAZ metal of wheel steel of type T decrease considerably under the impact of thermodeformational cycle of welding. In this connection it is anticipated that the HAZ metal of welded joints of the new wheel steel will feature a lower resistance to delayed cracking.

This value of the HAZ metal of higher strength wheel steel was evaluated by applying the universally known Implant method [5]. Unlike the traditional method the implants of 6 mm diameter from the studied steel were made without a notch [11]. Blanks from high-strength low-alloyed steel were used as the technological plates. The sample was inserted into a hole of the technological plate with a gap. Welding and loading of the samples were performed in a specialized unit, produced at PWI. During comparative testing, mechanized CO_2 welding with Sv-08G2S wire of 1.2 mm diameter was used in the modes ensuring the heat input on the level of 11.5 kJ/cm. Sample preheating temperature was varied from 20 up to 200 °C, this allowing adjustment of the HAZ metal cooling rate in the range of 25–5 °C/s.

Figure 3 gives the results of investigation of preheating temperature influence on critical fracture stresses of HAZ metal of new higher strength wheel steel. As is seen from the Figure, in welding without preheating (T_0 = 20 °C), when a predominantly martensitic component forms in the structure under the impact of the thermodeformational cycle ($w_{6/5}$ ~

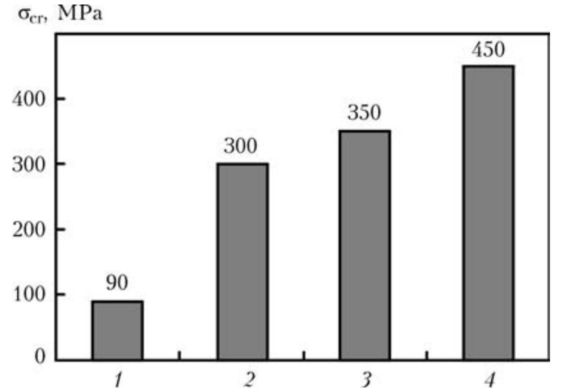

Figure 3. Influence of preheating temperature on delayed fracture resistance of HAZ metal of wheel steel of type T: *1* — 20; *2* — 70; *3* — 100; *4* — 150 °C

~ 25 °C/s) HAZ metal of wheel steel of type T has a low level of delayed fracture resistance. Critical fracture stresses σ_{cr} are equal to just about 90 MPa. Application of preheating promotes an essential increase of delayed fracture resistance. At preheating at $T_0 = 70$ °C, when the cooling rate is $w_{6/5} \approx 15$–18 °C/s and in the HAZ metal structure the overcooled austenite transformation occurs with formation of intermediate phases, values of critical fracture stresses increase up to 300 MPa. At T_0 = 100 °C ($w_{6/5} \approx$ ≈ 10–15 °C/s) σ_{cr} = 350 MPa. At T_0 = 150 °C ($w_{6/5} \approx$ ≈ 7–10 °C/s), when the HAZ metal forms structures with minimum content of the martensitic component, σ_{cr} values increase up to 450 MPa. In this case, critical fracture stresses are equal to approximately $0.45\sigma_y$ of HAZ metal or $0.40\sigma_t$ of wheel steel base metal. At increase of preheating temperature up to T_0 = 200 °C and higher ($w_{6/5}$ < 5 °C/s), when the overcooled austenite transformation in the HAZ metal runs solely with formation of ferrite-pearlite mixture, no delayed fracture of implants takes place.

Thus, analysis of thermokinetic diagram of overcooled austenite transformation of new higher strength wheel steel shows that under the impact of thermodeformational cycle of welding at cooling rate $w_{6/5} \geq 7.7$ °C/s a quenching structure forms in the HAZ metal of the new wheel steel, in which the martensite component fraction is higher than 60 vol.%. HAZ metal has high strength values ($\sigma_t \geq 1280$ MPa) and low ductility, which determines its increased susceptibility to delayed fracture at static loading. Under such cooling conditions the critical breaking stresses are not higher than $0.45\sigma_y$ of HAZ metal ($0.40\sigma_t$ of wheel steel).

In order to increase the delayed fracture resistance and prevent cold cracking in the joints (deposits) of higher strength wheel steel, it is necessary for cooling rate $w_{6/5}$ in the HAZ metal not to exceed 5 °C/s. Lowering of the cooling rate to the specified level is possible in the case of application of preheating to 200 °C and higher.

The given investigation results are the basic ones in development of technological recommendations on hardfacing the flanges of higher strength wheels. In order to achieve a high quality of the deposited metal and reliability of wheels in operation, it is necessary to conduct additional investigations on assessment of the technological factors on the strength of joints at static and cyclic loads. Such investigations are currently being conducted at PWI.

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