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INFLUENCE OF DEPOSITED METAL COMPOSITION ON STRUCTURE AND MECHANICAL PROPERTIES OF RECONDITIONED RAILWAY WHEELS

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Experimental data are given on the effect of composition of cladding consumables on formation of structure and mechanical properties of the deposited metal on wheels of steel 2. Strength properties, ductility and crack resistance of railway wheels repaired by cladding, were evaluated by analytical methods. It was found that to ensure the required combination of mechanical properties and high crack resistance of the base and deposited metals it is rational to apply cladding consumables of bainitic or bainitic-martensitic grade for repair of railway wheels of steel 2 by arc cladding.

Keywords: *arc welding, railway wheels, deposited metal, HAZ, structure, properties, crack resistance*

The problem of ensuring the reliability and fatigue life of rolling stock is becoming ever more urgent with increase of transportation volume and traffic intensity of railway transport. It is the most acute for basic parts and mechanisms of bogies of carriages and locomotives, the main part of which is the wheel, directly contacting the rail. Wheel tread wears in service. Flange inner surface is prone to considerable wear that is determined by service conditions of friction-rolling wheel–rail pair. Reconditioning of worn wheel tread is performed in specialized repair plants of railway transport by the method of mechanical turning, or by first performing repair cladding of flange surface.

Railway wheels of freight transport, locomotive wheel flanges, flanges of tram wheels of passenger transport are made of high-strength carbon steels, the composition and mechanical properties of which are given in Tables 1 and 2. As is seen, wheel steels feature high strength and hardness. Such metal properties ensure the required level of service strength of wheels. Flanges and all-rolled wheels from steel 2 are the most widely applied in railway transport of Ukraine and CIS countries. In keeping with GOST 10791–89, carbon content in wheel steel 2 is equal to 0.55–0.65 %. However, as shown by experience

Table 1. Composition of high-strength wheel steels, wt.%

Normative document	Steel grade	C	Mn	Si	V	S	P
						not more than	
GOST 10791–89	1	0.44–0.52	0.80–1.20	0.40–0.60	0.08–0.15	0.030	0.035
GOST 10791–89	2	0.55–0.65	0.50–0.90	0.22–0.45	≤0.10	0.030	0.035
TU U 35.2-23365425-600:2006	T	0.58–0.67	0.70–0.90	≤0.40	0.08–0.15	0.020	0.025

railway flanges and wheels are most often made from steel with not higher than 0.60 wt.% carbon content.

At present worn surfaces of railway wheels are repaired by cladding with application of several technologies, namely one- and two-wire submerged-arc cladding, and less often gas-shielded cladding. That is why cladding consumables used are solid welding wires of ferritic-pearlitic (Sv-08GS, Sv-08G2S), bainitic (Sv-10GSMT, Sv-08KhM) or bainitic-martensitic (Sv-10KhG23SMF) classes [1–3]. Worn flanges of all-rolled wheels of freight carriages are repaired by cladding with application of bainitic and bainitic-martensitic cladding consumables. Cladding is performed at controllable thermal cycle with application of preheating from 150 up to 250 °C and delayed cooling of wheels after cladding. These technologies ensure quality repair of worn metal of wheels and their performance during subsequent service. At the same time, at repair of the tread of tram wheel flanges 1960s technology is still used is some plants, when cladding was performed with Sv-08A wire without preheating or delayed cooling of items.

It is obvious that with such a diversity of technologies, the properties of deposited metal, fusion zone and HAZ metal regions of the joints differ essentially. This affects crack resistance of wheels repaired by cladding and their further serviceability. In terms of reliability and safety of railway traffic, during wheel repair by cladding it is necessary to ensure service properties of the deposited metal equivalent to those of wheel steel. For instance, at repair of flanges of all-rolled wheels of freight carriages, deposited metal hardness *HB* should be not lower than 2500 MPa, and ultimate strength — not lower than 700 MPa. It is necessary to ensure the homogeneity of the structure and minimum level of stress gradient in the zone of deposited-to-base metal transition, as item normalizing after cladding is not envisaged.

The objective of this work was performance of comparative evaluation of a set of deposited metal properties, depending on its composition, determination of peculiarities of structural changes in the deposits and their influence on mechanical properties of wheel steel joints. Wheel steel of grade 2 of the following composition (wt.%) was selected as the object of study: 0.58 C; 0.77 Mn; 0.44 Si.

Influence of thermodeformational cycle of welding on the structure and mechanical properties of wheel steel 2 is considered in [2], where it is shown that the cooling rate can have an essential influence on structural-phase composition and mechanical properties of HAZ metal. In the range of cooling rates $w_{6/5}$ = 1.15–32 °C/s ultimate strength of HAZ metal of wheel steel can change from 940 up to 1060 MPa, and yield point — from 600 up to 715 MPa. Its relative elongation δ_5 is in the range of 9.3–13.3 %, and reduction in area ψ — 24.9–33.3 %.

Table 2. Mechanical properties of high-strength wheel steels

Steel grade	σ_t, MPa	δ_5, %	ψ, %	KCU_{+20}, J/cm^2	*HB*, MPa
		not less than			
1	900–1100	12	21	30	2480
2	930–1130	8	14	20	2500
T	≥1100	8	14	18	3200

To determine deposited metal mechanical properties a method was selected, in which test specimens were cut out of upper layers of weld metal of multilayer welded joints with V-shaped groove of 350× 250×20 mm size, with 10 mm gap in the root. In such a variant of the joint mixing of base metal and weld in its central section is minimal, that can be equated to the conditions of multilayer cladding. Automatic submerged-arc welding of such joints was performed under a layer of AN-348 flux with solid wires of grades Sv-08G2S, Sv-08KhM and Sv-08KhMF of 2.0 mm diameter. Some joints were welded in CO_2 atmosphere with flux-cored wire PP-AN180MN (10KhGSMNFT alloying type) of 2.0 mm diameter. Welding of joints was performed with preheating up to the temperature of 150 °C. Welding modes were selected so as to ensure heat-input Q_w in the range of 11–13 kJ/cm. Furtheron special specimens for tensile (GOST 1497) and impact bend testing (GOST 9454) were cut out of upper layers of weld metal of the joints. Tables 3 and 4 give the composition and mechanical properties of weld metal. As shown by test results, except for Sv-08G2S wire, all cladding consumables allow producing deposited metal, the strength and hardness of which is on the level of requirements to reconditioned railway wheels of freight carriages.

As is known, deposited metal properties are determined by its structural-phase composition, which forms during welding and cooling of the joints. Therefore, the objective of further investigations was studying the influence of welding conditions on structural-phase composition of the metal of weld and HAZ of welded joints of wheel steel 2. Condition of metal structure was studied at various levels — from macro (grain) up to micro (dislocation) level.

Table 3. Weld metal composition, wt.%

Wire	C	Mn	Si	Cr	Ni	V	Mo
Sv-08G2S	0.10	2.10	0.95	–	–	–	–
Sv-08KhM	0.12	1.36	0.60	0.60	–	–	0.42
Sv-08KhMF	0.12	1.25	0.62	0.61	–	0.10	0.36
PP-AN180MN	0.12	1.00	0.35	0.67	0.80	0.10	0.40

Table 4. Mechanical properties of weld metal

Wire	$\sigma_{0.2}$, MPa	σ_t, MPa	δ_5, %	ψ, %	KCU_{+20}, J/cm^2	HB, MPa
Sv-08G2S	510	590	25.4	63.0	130	2100
Sv-08KhM	535	705	21.0	61.0	98	2500
Sv-08KhMF	610	730	17.2	56.3	86	2600
PP-AN180MN	700	790	12.7	57.6	97	2600

Allowing for these data and applying Archard equation, as well as well-known Hall–Petch, Orowan and other relationships, change of metal yield point $\sigma_{0.2}$ in different welded joint zones was determined by calculations:

$$\Sigma\sigma_y = \Delta\sigma_0 + \Delta\sigma_{s.s} + \Delta\sigma_{gr} + \Delta\sigma_s + \Delta\sigma_p + \Delta\sigma_d + \Delta\sigma_{d.s}, \qquad (1)$$

where $\Delta\sigma_0$ is the metal lattice resistance to free dislocation motion; $\Delta\sigma_{s.s}$ is the solid solution strengthening by alloying elements or impurities; $\Delta\sigma_{gr}$, $\Delta\sigma_s$ is the strengthening due to a change of grain and subgrain size; $\Delta\sigma_p$ is the strengthening due to pearlite; $\Delta\sigma_d$ is the dislocation strengthening; $\Delta\sigma_{d.s}$ is the dispersion hardening.

Contribution of individual structural components, size of grain, subgrain, dislocation density into the change of total (integral) strength was assessed. Method of conducting analytical evaluation at performance of these investigations is similar to that described in [4, 5].

Critical stress intensity factor K_{1C} characterizing brittle fracture resistance of the metal, and local internal stresses $\tau_{in.s}$ in the structure, which are the potential sources of crack initiation and propagation, were found from the following formulas:

$$K^*_{1C} = (2E\sigma_y\delta_t)^{1/2}, \qquad (2)$$

$$\tau_{in.s} = Gbh\rho/[\pi(1 - \nu)], \qquad (3)$$

where E is the Young's modulus; σ_y is the calculated value of yield point; δ_t is the value of critical crack opening displacement, equated to substructure parameters (subgrain size, lath width or fragment size); G is the shear modulus; b is the Burgers vector; $h = 2\cdot10^{-5}$ cm is the foil thickness; ν is the Poisson's ratio; ρ is the dislocation density.

All the above investigations were performed for cladding variants using solid wire Sv-08G2S and flux-cored wire PP-AN180MN, and their results are given in Figures 1–5.

As shown by metallographic investigations, weld metal of joints of wheel steel 2 made with Sv-08G2S wire has a coarse-grained structure, consisting of ferrite and pearlite (Figure 1, *a*).

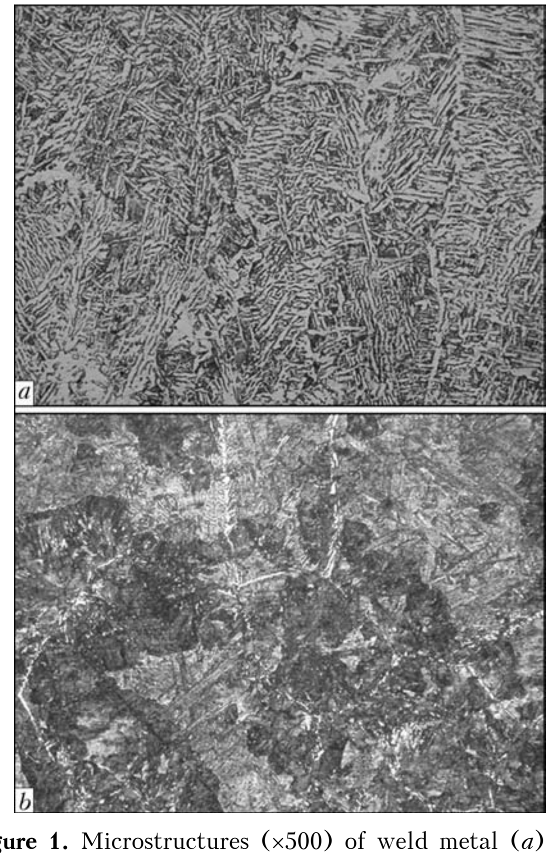

Figure 1. Microstructures (×500) of weld metal (*a*) and coarse-grain section of HAZ (*b*) of wheel steel 2 in welding with Sv-08G2S wire

Crystallite dimensions ($h_{cr}\times l_{cr}$) are tentatively equal to (30–130)×(60–250) μm. Ferrite microhardness is HV(F) = 2010, that of pearlite HV(P) = 2210–2510 MPa. Metal structure in the overheated zone of the HAZ of such joints consists of martensite (M), and bainite (B), which have the following microhardness, respectively: HV(M) = 3660–4540, HV(B) = 2570–3570 MPa, and ferrite fringes F_{fr} of thickness $\delta(F_{fr})$ = 3–7 μm, located along grain boundaries (Figure 1, *b*). Grain size D_{gr} of M and B is equal to 115–215 and 40–155 μm, respectively, and their volume fraction is V_M = 30 and V_B = 65–68 %.

Weld metal of joints made with flux-cored wire PP-AN180MN is characterized by bainite-martensite (B–M) structure (Figure 2, *a*) with microhardness HV(B–M) = 2860–3290 MPa and crystallite dimensions $h_{cr}\times l_{cr}$ = (25–60)×(65–180) μm, as well as presence of ferrite fringes of thickness $\delta(F_{fr})$ = 3–5 μm along grain boundaries. Ratio of B and M volume fraction in the weld metal is equal to 70–30 %. A structure close as to its composition was also found in the metal of HAZ overheated zone of such joints (Figure 2, *b*). B and M volume fraction in this welded joint zone is equal to 75–80 and 25–20 %, respectively, while microhardness is in the range of HV(M) = 3660–4540 MPa and HV(B) = 2570–3570 MPa. Sizes of B and M grains are equal to 115–215 (M) and 40–155(B) μm. As in the previous cases, ferrite fringes of thickness $\delta(F_{fr})$ = 3–7 μm were found along grain boundaries.

Thus, optical metallography showed that a coarse-grained ferrite-pearlite structure with an abrupt gradient of grain size (it differs by 2 to 4 times) forms in weld metal of joints made with

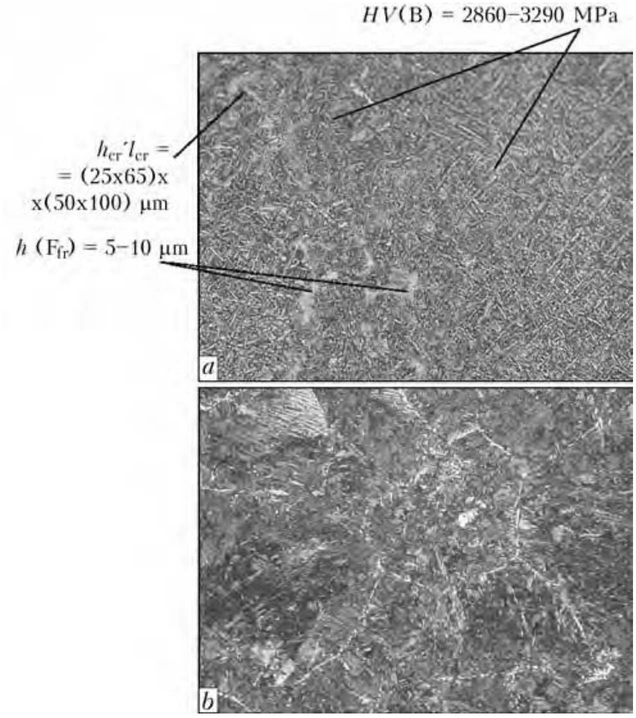

Figure 2. Microstructures (×500) of weld metal (*a*) and coarse-grain section of HAZ (*b*) of wheel steel 2 in welding with PP-AN180MN wire

Sv-08G2S wire, and a fine-grained bainite-martensite structure forms in the HAZ metal, its microhardness being practically 1.5 times higher than that of the deposited metal. In welded joints made with flux-cored wire PP-AN180MN, a bainite-martensite structure, quite close in its phase composition, and not having any abrupt gradients either by grain size, or by hardness, forms in the metal of weld and HAZ.

Features of variation of fine structure (lath width, substructure size and dislocation density) of the metal of weld and HAZ of wheel steel joints were studied by the methods of transmis-

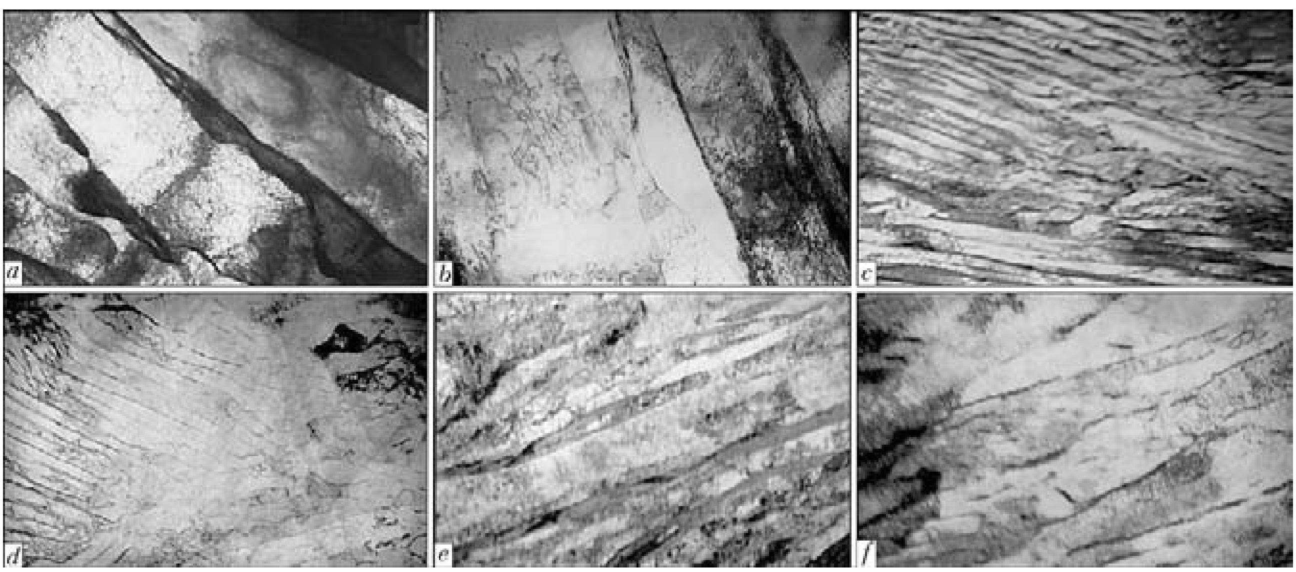

Figure 3. Fine structure of metal of weld (*a*–*d*) and coarse-grain section of HAZ (*e*, *f*) of joints of wheel steel 2 in welding with Sv-08G2S wire (*a*, *b*, *d*, *e* — ×20,000; *c*, *f* — ×30,000)

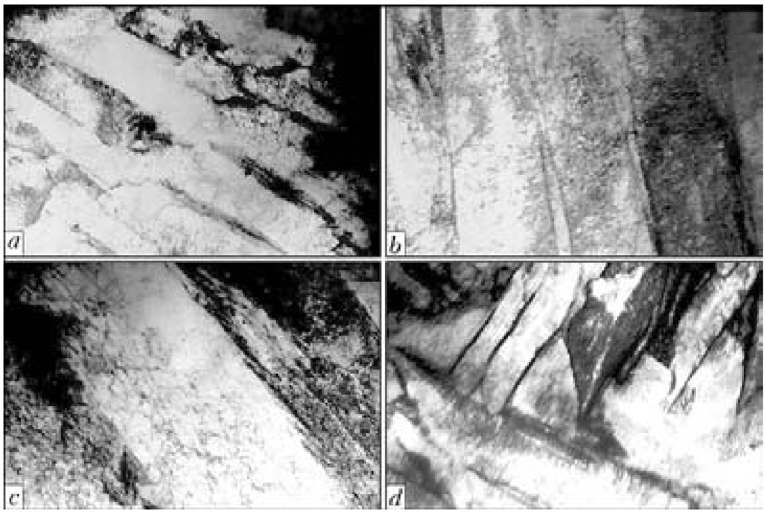

Figure 4. Fine structure of weld metal (*a*, *b*) and section of coarse-grain metal of HAZ (*c*, *d*) of joints of wheel steel 2 in welding with PP-AN180MN wire (*a* — ×15,000; *b* — ×20,000; *c*, *d* — ×30,000)

sion electron microscopy. They showed that in the main bulk of weld metal of joints made with Sv-08G2S wire, dimensions of cementite h_c and ferrite h_f plates in pearlite are equal to 0.1–0.4 and 0.7–1.5 μm, respectively, and dislocation density is $\rho = (4–6)\cdot10^9$ cm^{-2} (Figure 3, *a*, *b*). It is also established that in the weld section located in immediate proximity to the fusion line and at up to 500 μm distance from it, the structure is refined by an order of magnitude, and dislocation density rises up to $1\cdot10^{10}$ cm^{-2} (Figure 3, *c*, *d*). In the metal in HAZ overheated zone the width of upper bainite laths $h_{\rm Bup}$ is equal to 0.8–1.3 μm, that of lower bainite $h_{\rm Bl}$ is 0.3–0.8 μm, and of martensite $h_{\rm M}$ = 1.0–1.5 μm. Dimensions of fragments of lower bainite $d_{\rm fr}(\rm B_l)$ are in the range of 0.15–0.50 μm. Compared to weld metal, density of dislocations in this zone of welded joint rises up to $\rho = (5–8)\cdot10^{10}$ cm^{-2} (Figure 3, *e*, *f*).

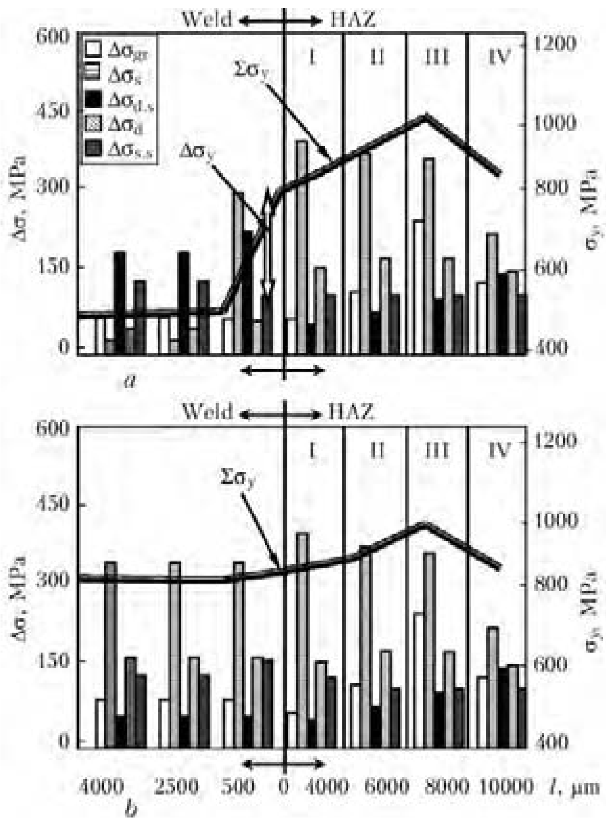

Figure 5. Contribution of structural components and strengthening of welded joint metal in welding with wires Sv-08G2S (*a*) and PP-AN180MN (*b*): I–IV — HAZ zones of overheating, normalizing, partial recrystallization and recrystallization, respectively

In welding with flux-cored wire PP-AN180MN a structure close as to its composition and element dimensions, consisting of upper and lower bainite, as well as martensite, forms in the metal of weld (Figure 4, *a*, *b*) and HAZ (Figure 4, *c*, *d*). Width of laths of upper bainite $h_{\rm Bup}$ is equal to 0.5–1.2 μm, that of lower bainite $h_{\rm Bl}$ = = 0.4–0.7 μm, and of martensite $h_{\rm M}$ = 1.0–1.5 μm. Distribution of dislocation density in such a joint is relatively uniform $\rho = 5\cdot(10^{10}–10^{11})$ cm^{-2}.

Figure 5 shows how metal yield point changes in different zones of welded joints, depending on deposited metal composition. It is established that strengthening of the bulk of weld metal of welded joints, made with Sv-08G2S wire, is due predominantly to the influence of carbide phases ($\Delta\sigma_{\rm d.s}$ = 190–230 MPa) (Figure 5, *a*). Near the fusion line weld metal yield point rises practically 1.7 times, from 480 up to 800 MPa. This occurs mainly due to additional contribution of substructural ($\Delta\sigma_{\rm s}$ = 300 MPa) and dislocation ($\Delta\sigma_{\rm d}$ = 60 MPa) factors into metal strengthening.

Change of the values of metal yield point in different zones of welded joints, made with flux-cored wire PP-AN180MN, occurs smoothly (Figure 5, *b*). Strengthening of weld metal of such joints occurs mainly due to substructural ($\Delta\sigma_{\rm s}$ = = 345 MPa) and dislocation ($\Delta\sigma_{\rm d}$ = 140–

200 MPa) strengthening. Carbide phase particles have a weaker effect on this process ($\Delta\sigma_{part}$ = = 75 MPa). Strengthening due to refining (dispersion) of lower bainite substructure is equal to about 33 % ($\Delta\sigma_s$ = 280 MPa) of the total strengthening level.

Strengthening of HAZ metal of the above-mentioned joints proceeds in a similar fashion. This is related, primarily, to fragmentation of lower bainite substructure ($\Delta\sigma_s$ = 367 MPa) and increase of dislocation density ($\Delta\sigma_d$ up to 200 MPa), that is equal to 42–48 and 20–25 % of the total strengthening level, respectively. Yield point of HAZ metal is in the range of 820–1000 MPa.

Results of calculation-based evaluation of K_{1C} values of welded joints of wheel steel 2 made with Sv-08G2S and PP-AN180MN wires (Figure 6) showed the following. Despite higher strength characteristics, the bulk of the metal deposited with flux-cored wire PP-AN180MN has the same K_{1C} values as the joints made with Sv-08G2S wire, and in some regions it exceeds them practically 2 times. This is associated with the features of formation of favourable substructure in them, more uniform distribution of dislocations and grain refinement, whereas metal deposited with Sv-08G2S wire features a non-uniform distribution of dislocation density and presence of pearlite structures with extended cementite phase precipitates that is detrimental to K_{1C} values. On the whole, such investigations showed that in welding with PP-AN180MN wire the weld metal forms a structure combining the high strength with good brittle fracture resistance.

Results of calculation-based evaluation of local internal stresses $\tau_{in.s}$ at comparison of these values with theoretical strength of material (Figure 7) showed that the lower general level of local internal stresses, distributed in the weld, forms in welded joints made with Sv-08G2S wire (Figure 7, *a*). Their value is not higher than 400 MPa, that is equal to approximately 0.04 of metal theoretical strength τ_{theor}. Increase of dislocation density from (4–6)$\cdot 10^9$ up to (5–8)$\cdot 10^{10}$ cm^{-2} in the weld zone, which is in immediate vicinity to the fusion line (FL) and in the HAZ of such joints, leads to an abrupt increase of $\tau_{in.s}$ up to 2240–2430 MPa, that is equal to (0.3–0.4)τ_{theor}, and to formation of considerable gradients of internal stresses ($\Delta\tau_{in.s}$ ~ 2000 MPa). Maximum $\tau_{in.s}$ values form on the mating boundary of F–P and B–M structures.

Total level of $\tau_{in.s}$ in weld metal in joints, made with flux-cored wire PP-AN180MN, is higher (1870–2240 MPa). They, however, are uniformly distributed in the weld metal and decrease relatively smoothly to 900–1100 MPa at transition to HAZ metal. Due to that no abrupt gradients of local stresses are observed in such joints that is favourable in terms of cracking prevention.

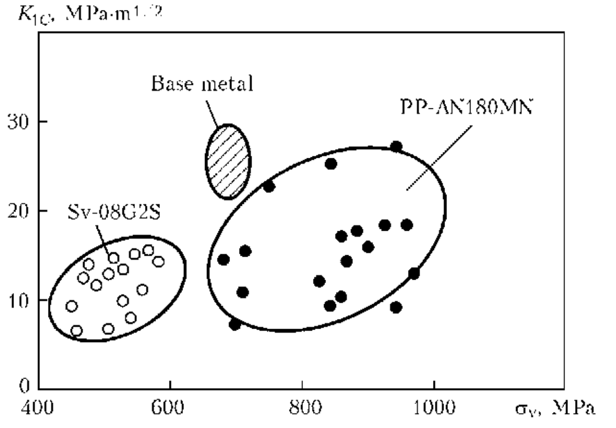

Figure 6. Diagram of the change of calculated mechanical properties ($\sigma_{0.2}$, K_{1C}) of weld metal of wheel steel welded joints depending on deposited metal composition

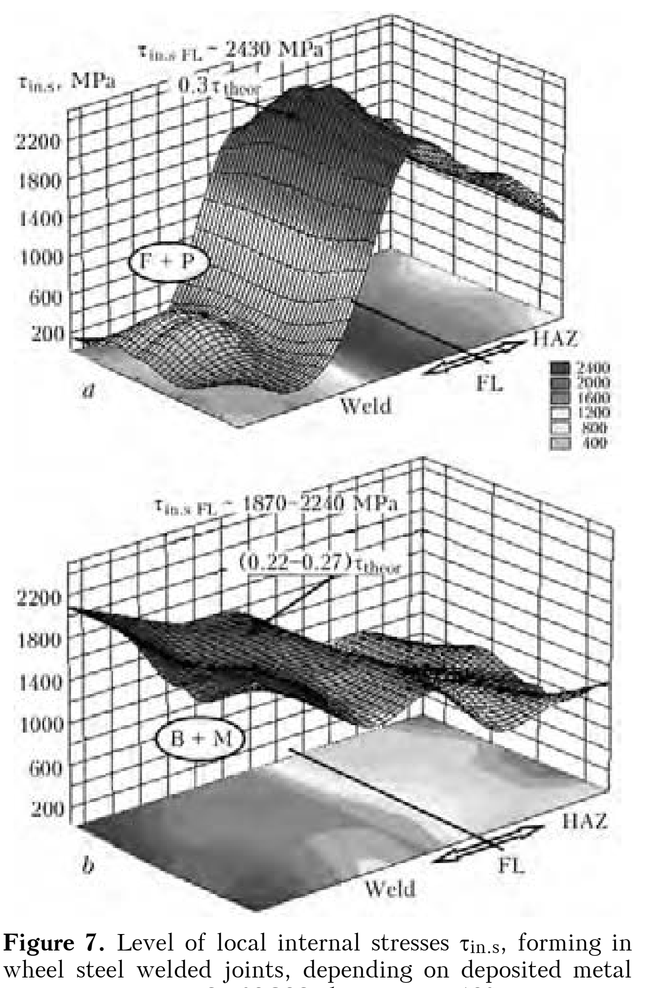

Figure 7. Level of local internal stresses $\tau_{in.s}$, forming in wheel steel welded joints, depending on deposited metal composition: *a* — Sv-08G2S; *b* — PP-AN180MN wire

CONCLUSIONS

1. Required set of mechanical properties of the deposited metal at reconditioning of worn surfaces of railway wheels (hardness $HB \geq 2500$, strength $\sigma_t \geq 700$ MPa) can be ensured by cladding consumables of bainite or bainite-martensite classes, namely solid wires Sv-08KhM, Sv-08KhMF and flux-cored wire PP-AN180MN.

2. Metal, deposited with flux-cored wire PP-AN180MN, combines high strength, hardness and crack resistance. All the regions of welded joints, made with this wire, have homogeneous finely-dispersed bainite-martensite structure with uniform distribution of local internal stresses.

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NEW SYSTEM OF FILLER METALS FOR BRAZING OF TITANIUM ALLOYS

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Based on the results of systematic studies of Ti–Zr–Co system alloys and allowing for published data, the liquidus surface of this system was plotted by using the simplex-lattice experimental design method. This system was found to have a region of alloys with decreased liquidus temperature, the most promising in terms of filler metal development. Spreading of experimental filler metals over titanium alloys of different classes (OT4, VT6 and VT22) was studied. The data on mechanical properties of the brazed joints are given.

Keywords: *brazing, titanium alloys, new system of brazing filler metals, liquidus surface, wetting, mechanical properties*

Titanium alloys play an important role in modern industry, especially in aircraft engineering, owing to their high characteristics, low density, high strength, high corrosion and fatigue resistance in particular, as well as high value of specific strength.

Meanwhile, the problem of current importance is fabrication of brazed structures of titanium alloys which cannot be manufactured by welding. Such structures include critical heat exchangers for cooling of compact nuclear reactors, as well as lamellar-ribbed and honeycomb structures used in aircraft engineering, ship building, etc. [1].

The modern brazing technology and filler metals should provide seams with the properties close to those of the base metal. This specifies important temperature-time bounds of the brazing cycles determined by the nature of titanium alloys. These bounds limit the probability of undesired changes in structure and properties of the alloys caused by polymorphism of titanium.

At a temperature below 882 °C titanium is in the α-state (hexagonal lattice), while above this temperature it is in the β-state (cubic lattice). This circumstance has a considerable effect on diffusion of depressant elements from the seam to a metal brazed and, as a consequence, on structure and properties of the brazed joints.

According to study [2], the need to limit the brazing temperature of titanium and its alloys is caused by a high rate of growth of its grain at temperatures above 1000 °C. Therefore, the melting point of a filler metal should not exceed 950–1000 °C.

In study [3] the upper bound of the brazing temperature was decreased to 900 °C for α- and pseudo-α-alloys, to 935 °C for $\alpha + \beta$-alloys, to 870 °C for pseudo-β-alloys, and to 760–800 °C for β-alloys.

Limitation of the brazing temperature to a value of the $\alpha \rightarrow \beta$ transformation temperature is especially important for thin-walled structures