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STRUCTURE AND WEAR RESISTANCE OF CHROME-MANGANESE DEPOSITED METAL

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Presented are the results of investigations into structure and wear resistance of low-carbon metal of different structural classes, deposited with flux-cored strips and containing approximately 13 % chromium and different amounts of manganese (from 2 to 12 %). The possibility of increasing wear resistance of the deposited metal by tempering and case hardening after hardfacing was studied. It is shown that achieving the optimal amount of meta-stable austenite, along with martensite, in structure of the deposited metal improves its wear resistance.

Keywords: arc hardfacing, chrome-manganese deposited metal, structure, martensite, meta-stable austenite, tempering, case hardening, wear resistance

To repair and strengthen parts operating under conditions of a mechanical wear combined with the corrosion effect at normal and increased temperatures, the industry widely applies hardfacing consumables that provide deposited metal of the type of low-carbon high-chromium steels (≤ 0.2 % C, ~ 13 % Cr) of the martensitic-ferritic grade. They are used to repair and strengthen plungers of hydraulic presses and hydraulic cylinders, continuous casting machine rollers, components of power and petrochemical fixtures, etc. [1].

Ferrite is characterised by the lowest fracture resistance, compared to martensite and austenite, which hampers achieving a high wear resistance of the deposited metal. To decrease the content of ferrite and provide mostly the martensitic structure, the deposited metal is alloyed with nickel in an amount of 2–4 %. Examples of such hardfacing consumables are PP-Np-12Kh13N2MFA and PP-Np-12Kh14N3. Along with nickel, formation of ferrite is also suppressed by

alloying with less expensive manganese. Properties of the low-carbon deposited metal with approximately 13 % Cr and different contents of manganese have been insufficiently studied as yet. However, the data on structure and properties of this type of chrome-manganese steels can be found in studies [2–4].

The purpose of this study was to investigate structure of the low-carbon Cr–Mn deposited metal of different structural grades (martensitic, martensitic-austenitic, austenitic-martensitic, austenitic) and determine its wear resistance under different test conditions to define rational compositions of hardfacing consumables for different service conditions. The possibility of improving properties of the deposited metal due to tempering parameters and case hardening was also investigated.

One-lock flux-cored strips with a cross section of 10×3 mm and a fill factor of 48–50 % were manufactured for hardfacing of experimental samples. Cold-rolled strip of steel 08kp (rimmed) was used as a steel sheath. Manganese and chromium metals, iron powder and a small amount of ferrotitanium to refine grains and strengthen the deposited metal due to formation of dispersed carbides TiC were added in different quantities to the charge composition. Hardfacing was performed in three layers by the submerged-arc method using flux AN-26 on a 30 mm thick plate of steel VSt3sp (killed). The hardfacing parameters were as follows: current 450–500 A, voltage 30–32 V, and hardfacing speed 25 m/h. Chemical compositions of the metal deposited with wire Sv-12Kh13 and experimental flux-cored strips PL-OP (1–5) are given in Table 1.

As found at the preliminary stage of investigations, in hardfacing without preheating the deposited metal (~ 13 % Cr and ≥ 6 % Mn) was free from cracks. No cracks were detected in the metal deposited with wire Sv-12Kh13 either. However, they were detected at 2–4 % Mn. Preheating to a temperature of 250 °C made it possible to avoid cracks in all the cases. Al-

Table 1. Chemical composition of deposited metal, %

Hardfacing consumable	C	Cr	Mn	Si	Ti
Sv-12Kh13	0.11	12.6	0.7	0.55	–
PL-OP1	0.12	12.5	2.3	0.73	0.14
PL-OP2	0.15	13.3	4.1	0.62	0.17
PL-OP3	0.13	13.1	6.2	0.71	0.15
PL-OP4	0.16	12.9	7.8	0.65	0.18
PL-OP5	0.17	12.6	12.2	0.66	0.16

Note. Content of S and P ≤ 0.03 wt.%.



lowing for this fact, preheating was used only when necessary in hardfacing of samples for further investigations.

The effect of the tempering temperature, as well as of case hardening on wear resistance of the deposited metal with different manganese contents was also investigated. Tempering of the hardfaced samples was carried out at temperatures of 450, 550, 650 and 750 °C with holding for 1 h and subsequent cooling in air. Case hardening of the samples was performed in a solid carbonaceous medium at a temperature of 950 °C for 10 h, followed by cooling of the samples in air. After case hardening, some of the samples were subjected to tempering at a temperature of 650 °C.

Relative wear resistance of the deposited metal under conditions of dry friction by the block-roller scheme was evaluated at room temperature (ϵ_1) and at 500 °C (ϵ_2). The 70 mm diameter roller used for wear was hardfaced by the TIG welding method using alloy P18 (*HRC* 56). The roller rotation speed was 200 rpm. The sample was pressed to the roller by the 8 kg load. Wear at the increased temperature was evaluated by using a continuous furnace working chamber after preliminary holding of the sample for 15 min.

Relative wear resistance ϵ_3 was determined by affecting the surface of a flat sample by a grit transported with a flow of compressed air under a pressure in the system equal to 5 atm. Diameter of the nozzle at exit of the jet from a mixer was 16 mm. Duration of the tests of the samples was determined by a grit consumption equal to 20 kg. The relative wear resistance was evaluated from a ratio of the losses of weights of a reference and the sample related to the area of their worn out surfaces, respectively. This test method used to evaluate resistance to fatigue fracture of metal under the repeated force effect is considered in study [5].

One of the main causes of failure of the continuous casting machine rollers is thermal cracking [1]. Resistance to thermal cracking (fire cracking) of the metal deposited with the experimental flux-cored strips was determined from the quantity of heating-cooling cycles to appearance of the cracks that could be seen with unaided eye. The fire crack resistance tests were conducted according to the procedure similar to that suggested in study [6], the difference being that the samples were heated not in a melt of salts, but in a laboratory furnace at a temperature of 550 °C for 15 min, after which they were cooled in water.

Note that in all the cases the reference for evaluation of properties of the deposited metal was metal deposited with wire Sv-12Kh13.

Durometric and metallographic examinations were carried out. Phase composition was determined by the X-ray method using diffractometer DRON-4. Variations in the amount of the magnetic phase in the deposited metal over a range of 5 to 60 % C were estimated by using ferritometer FM-1.

Results of the tests to wear resistance, hardness and content of the ferritic component after hardfacing without tempering, as well as at different temperatures of tempering, are given in Table 2.

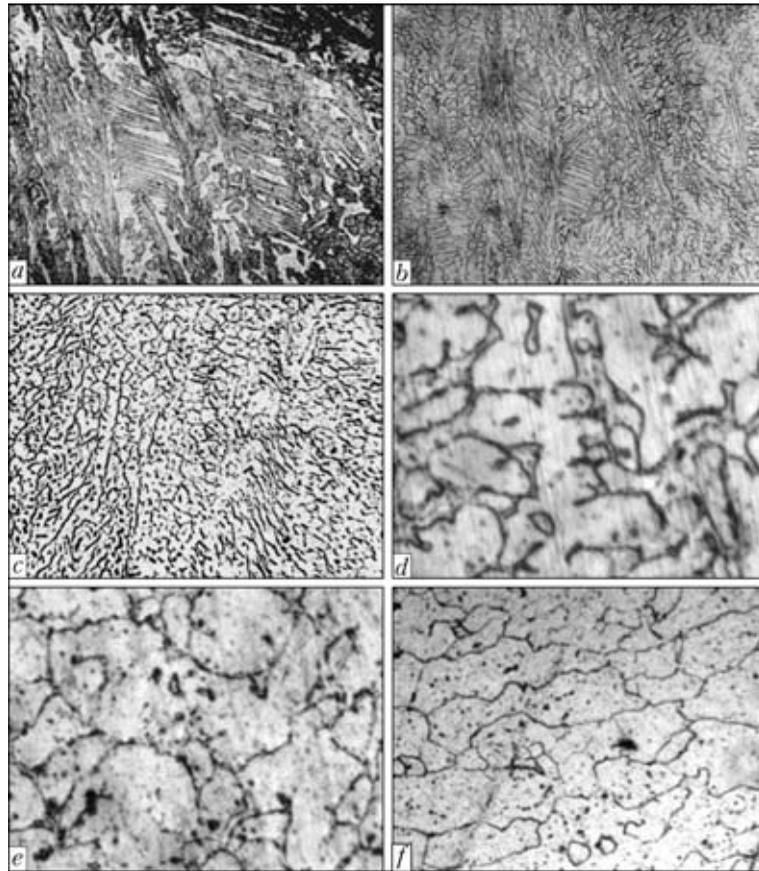
Microstructures of the deposited metal with different contents of manganese after hardfacing without tempering are shown in the Figure.

The metal deposited with wire Sv-12Kh13 and containing 0.7 % Mn has a martensitic-ferritic structure (Figure, *a*). Its hardness is *HRC* 36–38. At approximately 2 % Mn, structure of the deposited metal is mostly martensitic. Along with martensite, also a small amount (≤ 5 %) of high-temperature ferrite may persist in structure. Hardness of the deposited metal is *HRC* 40–42. As the manganese content is increased to 4 %, retained austenite appears in structure of the

Table 2. Effect of manganese and tempering temperature on properties of deposited metal

Hardfacing consumable	Tempering temperature, °C	Hardness <i>HRC</i>	Content of ferromagnetic component, %*	Relative wear resistance	
				ϵ_1	ϵ_3
Sv-12Kh13	Without tempering	36–38	100	1.0	1.0
	450	34–36	100	1.1	1.0
	550	31–33	100	0.8	0.8
	650	23–25	100	0.6	0.7
	750	34–36	100	0.9	1.1
	750	34–36	100	100	0.9
PL-OP1	Without tempering	40–42	100	1.6	1.2
	450	41–43	100	1.8	1.3
	550	38–40	100	1.4	1.1
	650	36–38	100	0.9	0.9
	750	40–42	100	1.5	1.3
	750	40–42	100	100	1.5
PL-OP2	Without tempering	38–40	80–85	1.8	1.5
	450	40–42	80–85	1.9	1.7
	550	37–39	85–90	1.6	1.4
	650	34–36	90–95	1.2	1.2
	750	39–41	75–80	1.7	1.5
	750	39–41	75–80	100	1.7
PL-OP3	Without tempering	35–37	65–70	1.6	1.8
	450	36–38	65–70	1.7	2.0
	550	35–37	70–75	1.5	1.7
	650	34–36	75–80	1.4	1.6
	750	36–38	60–65	1.5	1.8
	750	36–38	60–65	100	1.5
PL-OP4	Without tempering	32–34	25–30	1.4	2.2
	450	30–32	25–30	1.2	2.5
	550	33–35	30–35	1.4	2.3
	650	34–36	35–40	1.6	1.9
	750	32–34	20–25	1.3	2.1
	750	32–34	20–25	100	1.3
PL-OP5	Without tempering	22–24	5–10	1.2	2.0
	450	20–22	5–10	1.1	1.9
	550	23–25	10–15	1.2	2.1
	650	24–26	15–20	1.4	2.2
	750	22–24	3–5	1.2	2.0
	750	22–24	3–5	100	1.2

*Content of the carbide phase is less than 1 %, balance – austenite.



Microstructures ($\times 550$) of Cr-Mn deposited metal with different manganese contents: *a* – Sv-12Kh13 (reference); *b* – 2; *c* – 4; *d*, *e* – 8; *f* – 12 % Mn

deposited metal. With further increase in the manganese content of the deposited metal the amount of austenite grows, and hardness of the deposited metal decreases (see Table 2).

Under conditions of dry friction at an ambient temperature, the deposited metal containing 2 % Mn has a 1.6 times higher relative wear resistance ϵ_1 than the Sv-12Kh13 deposited metal. The maximal wear resistance ($\epsilon_1 = 1.8$) is achieved at 4 % Mn, which is caused by solid-solution strengthening of martensite alloyed with manganese, as well as by dynamic deformation martensite transformation (DDMT) of meta-stable retained austenite taking place in wear. The growth of the latter on the wearing surface is 10–17 %. As the manganese content of the deposited metal is increased to 6 %, the amount of retained

austenite grows to 30–35 %. This decreases to some extent the relative wear resistance ($\epsilon_1 = 1.6$). Its lowest level corresponds to 12 % Mn ($\epsilon_1 = 1.2$), when the deposited metal structure becomes mostly austenitic. This can be explained by the fact that the growth of martensite on the wearing surface, which is 30–35 %, cannot compensate for the almost absolute absence of cooling martensite.

When wear occurs in the grit flow, with increase of the manganese content from 2 to 8 % the relative wear resistance grows more than 2 times (to $\epsilon_3 = 2.2$), compared to the reference, and remains at a high level ($\epsilon_3 = 2.0$) at 12 % Mn. This is attributable to the fact that deformation of the wearing surface by grit is much higher than in dry friction, this leading to a more intensive development of DDMT. The growth of martensite on the wearing surface of the sample at 8 % Mn in the deposited metal is 45–50 %, and at 12 % Mn – 38–42 %. This causes substantial strengthening of the surface of the deposited metal (hardness in this case grows to *HRC* 45–50) and increase in fracture resistance. A considerable part of energy of the abrasive particles is consumed for DDMT, which is characterised by an intensive development [7].

The data on hardness, relative wear resistance ϵ_2 and fire crack resistance of the deposited metal at 550 °C depending on its manganese content are given

Table 3. Fire crack resistance and wear resistance of deposited metal under dry friction conditions at 550 °C

Hardfacing consumable	Hardness <i>HRC</i>	Fire crack resistance	Relative wear resistance ϵ_2
Sv-12Kh13	30–32	1.0	1.0
PL-OP1	38–40	0.8	1.7
PL-OP2	36–38	0.9	2.0
PL-OP3	33–35	1.1	1.8
PL-OP4	30–32	1.2	1.5
PL-OP5	20–22	1.0	1.3



in Table 3. Hardness at 550 °C decreases with increase of the manganese content. Moreover, decrease in hardness of the deposited metal, compared to its level at an ambient temperature, is fixed for the corresponding manganese contents.

Dependence of relative wear resistance ε_2 of the deposited metal on its manganese content (see Table 3) is similar to that considered above for ε_1 determined at the ambient temperature. However, the values of wear resistance ε_2 for similar compositions of the deposited metal are higher than those of ε_1 . Increase in relative wear resistance ε_2 can also be caused by strengthening of martensite alloyed with manganese, as well as by the effect of dynamic ageing of martensite taking place in wear.

The lowest fire crack resistance (see Table 3) is exhibited by the deposited metal containing about 2 % Mn with mostly the martensitic structure. Increase of the manganese content to 8 %, when structure becomes austenitic-martensitic, increases the fire crack resistance, whereas further increase of the manganese content to 12 % causes its decrease. The given regularity is explained by the fact that increase in the content of austenite, which is a more ductile phase than martensite, leads to increase in crack resistance. Decrease in fire crack resistance with increase in the content of austenite above its optimal amount is related to the fact that the γ -phase has an almost twice as high thermal expansion coefficient, compared to the α -phase, this leading to a growth of stresses forming in structure.

The effect of the tempering temperature on properties of the deposited metal was investigated. It was found to depend on the proportion of martensite and austenite in the initial structure after hardfacing. The common pattern of variations in properties is observed for the deposited metal with mostly the martensitic structure at 2–6 % Mn. The highest hardness and relative wear resistance under dry friction conditions, ε_1 , and in the grit flow, ε_3 , are achieved after tempering at a temperature of 450 °C, this being provided by precipitation hardening and activation of DDMT [4]. With increase of the tempering temperature from 450 to 650 °C, hardness and relative wear resistance ε_1 and ε_3 decrease due to weakening of martensite. As the tempering temperature is further increased to 750 °C, again this causes a rise in hardness and relative wear resistance ε_1 and ε_3 . This is a result of heating into the inter-critical temperature range ($A_{c3} > T > A_{c1}$) causing redistribution of carbon between the α - and γ -phases, thus leading in cooling to formation of martensite with the increased carbon content from the enriched γ -phase.

For the deposited metal having mostly the austenitic structure, at 8 and 12 % Mn the lowest values of hardness and relative wear resistance in dry friction, ε_1 , are fixed after tempering at a temperature of

450 °C. This is caused by increase in stability of austenite due to blocking of the dislocations providing an increase in energy required for DDMT. Raising the tempering temperature to 650 °C increases hardness and relative wear resistance ε_1 up to the maximal values because of precipitation of carbides and activation of DDMT [8]. After tempering at 750 °C, the values of hardness and relative wear resistance ε_1 decrease due to dissolution of part of carbides, thus increasing the amount and stability of austenite.

At 8 and 12 % Mn in the deposited metal the tempering temperature was found to have a differing effect on relative wear resistance in the grit flow, ε_3 , which is caused by a differing stability of austenite. Tempering at 450 °C increasing stability of austenite raises relative wear resistance ε_3 in the case when stability of austenite is low (at 8 % Mn). And on the contrary, tempering at 650 °C destabilising austenite increases wear resistance ε_3 in the case when stability of austenite is increased (at 12 % Mn).

The investigations performed showed the expediency of application of the flux-cored strips, which provide the low-carbon Cr–Mn deposited metal with 4–6 % Mn (with martensitic-austenitic structure), for hardfacing of parts operating under dry friction conditions. They are characterised by indices 19, 25, 26 and 34, according to the classification of study [9].

Hardfacing flux-cored strips, which provide the deposited metal containing 8 % Mn and having the austenitic-martensitic structure, are efficient for the use for hardfacing of parts operating under contact-fatigue loading (corresponding to indices 18, 23, 24, 30, 35 [9]). At a not less than 6 % Mn content of the deposited metal it is possible to perform hardfacing without preheating, this being a technological advantage and reducing the power intensity of the process [10].

The deposited metal containing 12 % Mn has the austenitic structure, which provides its good manufacturability (workability by cutting, low sensitivity to cracking) and resistance to the contact-dynamic effect. However, its relative wear resistance under dry friction conditions and crushing resistance are insufficient. Hardfacing with the flux-cored wire providing the 12 % Mn content can be recommended for parts operating under conditions characterised by indices 21 and 27–29 [9].

Increasing the carbon content of the deposited metal improves its relative wear resistance, but hampers machining and often leads to cracking. The technique that makes it possible to handle this contradiction is hardfacing with low-carbon steels followed by their case hardening [11]. This technique can be effectively used for strengthening and repair of parts operating under the conditions characterised by indices 15, 25 and 26, e.g. necks of axles and shafts, contact surfaces of hinged joints, etc. The expediency of case hardening of the deposited metal is proved in study [12].

After case hardening, carbides of the Cr_7C_3 type form in a surface layer of the Cr–Mn deposited metal

**Table 4.** Properties of deposited metal after case hardening and subsequent tempering

Hardfacing consumable	Treatment	Structure	Hardness HRC	Relative wear resistance	
				ϵ_1	ϵ_3
Sv-12Kh13	Case hardening	15–17 % A; M + C – balance	45	1.7	1.2
	Same + tempering at 650 °C	10–12 % A; FCM + C – balance	42	1.4	1.1
PL-OP1	Case hardening	20–25 % A; M + C – balance	52	2.5	1.5
	Same + tempering at 650 °C	15–18 % A; FCM + C – balance	48	2.2	1.3
PL-OP2	Case hardening	35–40 % A; M + C – balance	55	4.3	1.9
	Same + tempering at 650 °C	28–30 % A; FCM + C – balance	50	4.0	1.5
PL-OP3	Case hardening	55–60 % A; M + C – balance	44	3.8	2.4
	Same + tempering at 650 °C	45–50 % A; FCM + M' + C – balance	48	4.2	2.1
PL-OP4	Case hardening	75–80 % A; M + C – balance	38	3.0	2.5
	Same + tempering at 650 °C	65–70 % A; FCM + M' + C – balance	41	3.4	2.8
PL-OP4	Case hardening	80–85% A; C – balance	35	2.6	2.2
	Same + tempering at 650 °C	75–80 % A; M' + C – balance	37	2.9	2.5

Note. A – austenite; C – carbide; FCM – ferrite-carbide mixture; M – martensite; M' – martensite formed in cooling from tempering temperature.

with 13 % Cr. In all the cases the matrix consists of austenite and martensite with different proportions of these phases. As the manganese content is increased, the amount of austenite grows and that of martensite decreases accordingly (Table 4). The carbon content of the surface layer was about 1.6 %, and thickness of the case hardened layer was 1.3–1.5 mm.

High stability of the investigated compositions of the deposited metal subjected to case hardening, with respect to decomposition of overcooled austenite to form the ferrite-carbide mixture, allows refusing from hardening in a liquid coolant, as self-hardening occurs in air cooling. Important in this case is reduction of stresses and residual strains in parts, simplification and environment friendliness of the technological process.

The Cr–Mn deposited metal with 4 % Mn has the highest wear resistance after case hardening under dry friction conditions. In this case about 40–50 % of meta-stable austenite is fixed in the structure, along with martensite and carbides. As the manganese content is further increased, wear resistance of the deposited metal decreases due to increase in the amount and stability of austenite in its structure. Subjecting the deposited metal with maximum 4 % Mn to tempering at 650 °C leads to deterioration of its wear resistance ϵ_1 . This is caused by weakening of martensite and decomposition of retained austenite. And on the contrary, relative wear resistance ϵ_1 of the deposited metal containing not less than 6 % Mn grows because of activation of DDMT due to precipitation of carbides from austenite. These data are in agreement with the results of study [13].

In testing in the grit flow, the highest relative wear resistance ϵ_3 was achieved after case hardening

of the deposited metal containing 8 % Mn and having 75–80 % of retained austenite in its structure. Tempering at a temperature of 650 °C has a differing effect on wear resistance in the grit flow, ϵ_3 , for the deposited metal with differing manganese contents: at not higher than 6 % Mn it decreases, and at not less than 8 % Mn it increases. This is attributable to the fact that at no more than 6 % Mn tempering at 650 °C leads to weakening of martensite and excessive activation of DDMT, where stability of austenite is low, this leading to decrease in relative wear resistance ϵ_3 . At not less than 8 % Mn, where stability of austenite is increased, on the contrary the activation of DDMT raises relative wear resistance ϵ_3 .

The method suggested for improving wear resistance by hardfacing with the low-carbon Cr–Mn-based metal, followed by case hardening [11], in all the cases provides formation of retained meta-stable austenite in structure, along with martensite and carbides. However, it should be taken into account that the highest wear resistance for a specific type of a test is achieved at the optimal amount and stability of austenite, which can be controlled by tempering.

CONCLUSIONS

1. It is shown that development of hardfacing consumables providing the deposited Cr–Mn metal of different structural grades and having the martensitic, martensitic-austenitic, austenitic-martensitic or austenitic structure, depending on the imposed requirements, holds high promise for extending the life of parts operating under conditions of dry friction and contact-fatigue loading.

2. The amount and stability of austenite with respect to DDMT have a substantial effect on wear



resistance, and should be controlled by tempering parameters, allowing for the specific loading conditions.

3. The efficiency is shown of improving wear resistance by combining hardfacing with the low-carbon Cr–Mn-based metal and subsequent case hardening.

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PECULIARITIES OF STRUCTURE AND MECHANICAL HETEROGENEITY IN EB-WELDED JOINTS OF 1201-T ALLOY

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The peculiarities of microstructure and distribution of microhardness of weld and heat-affected zone metal of welded joints of 1201-T alloy, produced using electron beam welding at different values of heat input, were studied. Using plotted temperature fields the running of phase transformations occurred in welding was analyzed. It was established that due to change of cooling rate across the thickness of plates the microstructure of near-weld zone in upper and lower parts of welded joint were considerably different. The increase of heat input of welding facilitates recrystallization processes which leads to increase of heat-affected zone.

Keywords: *electron beam welding, aluminium alloy, heat-affected zone, weakening, temperature field, structure and mechanical heterogeneity, microstructure*

Due to high strength and resistance to cryogenic and corrosion embrittlement the aluminium alloys of Al–Cu–Mn system are widely applied in aircraft industry [1, 2]. For joining critical elements of structures the electron beam welding (EBW) is applied as far as this method provides high quality of weld metal in one-pass welding large thicknesses [3]. The characteristic property of thermal cycle of welding is rapid heating of metal and also its cooling. The short periods of this process provide special kinetics of structural transformations which results in weakening and heterogeneity of welded joints (WJ) [4–12]. It is the most clearly revealed in welding of thick plates and caused by decay of solid solution of copper in aluminium and partial coagulation of strengthening phases due to uniform heating of weld and adjacent zones of base metal [4].

Elements of structures of aluminium alloys containing WJ are operating under conditions of alter-

nating dynamic loads, deep vacuum and cryogenic temperatures, which frequently results in initiation and propagation of micro- and macrofracture in them. The process of microcrack propagation depends directly on the structure and mechanical characteristics of metal. Therefore, to perform efficient diagnostics of WJ of structure elements manufactured of mentioned aluminium alloy including methods of acoustic emission [5–7], it is important also to investigate the microstructure of HAZ metal.

The purpose of the work is to study peculiarities of structure and mechanical heterogeneity of WJ of thick plates of heat-hardened 1201-T alloy, produced using EBW.

It is known that level of weakening and structure changes in HAZ metal of WJ of heat-hardened aluminium alloys is determined by welding cycle and structure of base metal [8–10]. It was also established [11] that a weld is weakened to the level characteristic for metal in annealed state (for 1201 alloy to HRB 70). This phenomenon is predetermined by both the processes of dissolution of strengthening phases as well as their further precipitation during cooling.