STRUCTURE AND PROPERTIES OF DEPOSITED WEAR-RESISTANT Fe–Cr–Mn STEEL WITH A CONTROLLABLE CONTENT OF METASTABLE AUSTENITE

Ya.A. CHEJLYAKH and **V.V. CHIGAREV**
Priazovsky State Technical University, Mariupol, Ukraine

The paper gives the results of investigation of the structure and phase transformations of metal deposited with flux-cored wire PP-Np-20Kh12G10SF. Possibility of controlling the content, degree of metastability of austenite and wear resistance of deposited metal of Fe–Cr–Mn system is shown.

Keywords: *arc hardfacing, flux-cored wire, wear resistance, metastable austenite, martensite*

Development of highly efficient methods of reconditioning and strengthening of worn working surfaces of rapidly wearing parts of diverse equipment remains to be a highly urgent task. One of its solutions is development of hardfacing materials, providing deformational metastability of austenite of the deposited alloyed steel, capable of considerable deformation strengthening during wear [1–3]. A considerable contribution into formation of properties of such deposited metal is made by realization of deformation $\gamma \rightarrow \alpha'$ and $\gamma_{res} \rightarrow \alpha'$ martensite transformations at testing (DMTT) or in service [2–8]. However, these advantages are so far insufficiently used in flux-cored hardfacing consumables and metastable wear-resistant steels deposited using them, and published data on studying the capabilities of controlling the degree of metastability of deposited metal are quite limited [6, 7]. Therefore, development of new electrode materials, ensuring deposition of such metastable sparsely-alloyed steels with a controllable amount and metastability of austenite, is a quite urgent task, which is of a certain scientific and applied interest.

The purpose of this work is studying the structure, phase transformations, possibilities for controlling the quantity and metastability of austenite and properties of Fe–Cr–Mn wear-resistant steel deposited with developed flux-cored wire PP-Np-20Kh12G10SF [8].

Deposition with developed 4 mm flux-cored wire was performed with AN-348 flux on steel St3 and steel 45 in the following modes: I_{hf} = 320–480 A, U_a = 28–32 V, v_{hf} = 22–32 m/h. Various design-technological schematics of hardfacing were used: in one, two and three layers, as well as four-five layers similar to the way it is done for actual parts. Samples for investigations were prepared from the deposited metal. Fraction of base metal in the weld metal was determined by the ratio of penetration area to total area of deposited metal. Digital images were entered into the computer, where they were processed by the procedure from [9]. Fraction of base metal was varied by adjustment of hardfacing parameters (I_{hf}, U_a, v_{hf}) that allowed controlling the composition of the deposited metal, particularly chromium and manganese content [1].

Analysis of chemical composition was conducted in vacuum quantometers «Spectrovac 1000» and «SpectroMAXx» by the spectral method, phase analysis of deposited metal was performed in X-ray diffractometer DRON-3 in Fe-K_α-radiation, in the range of angles $2\theta = 54–58°$, macroscopic analysis — on transverse and longitudinal macrosections of deposited metal with deep etching in an acid mixture (100 ml HCl, 10 ml HNO_3 and 100 ml H_2O). Microstructures were studied in metallographic microscopes MMR-2 and «Neophot-21» (×50–500), microhardness of structural components was measured in PMT-3 microhardness meter by indentation of a tetrahedral diamond pyramid under 1.96 N load, and deposited metal hardness — in TK (Rockwell) instrument with 1500 N (*HRC*) and 600 N (*HRA*) load. Dynamic testing was conducted in an impact pendulum-type testing machine IO5003 on samples of 10 × 10 × 55 mm size with U-shaped notch. Wear testing at dry metal-to-metal friction was performed in MI-1M machine on samples of 10 × 10 × 27 mm size by the schematic of the block (tested sample)–roller (rider) rotating at the speed of 500 min^{-1} (linear speed in the friction zone of 1.31 m/s, friction path of 1965 m). Time of wearing between two weighing operations was equal to 5 min that ensured heating of the contact surface (similar to heating of equipment parts under the actual wearing conditions), total wearing time was 25 min. Weighing was conducted with the accuracy of up to 0.0001 g. Relative wear resistance was determined from the following formula

$$\varepsilon = \frac{\Delta m_r}{\Delta m_s},$$

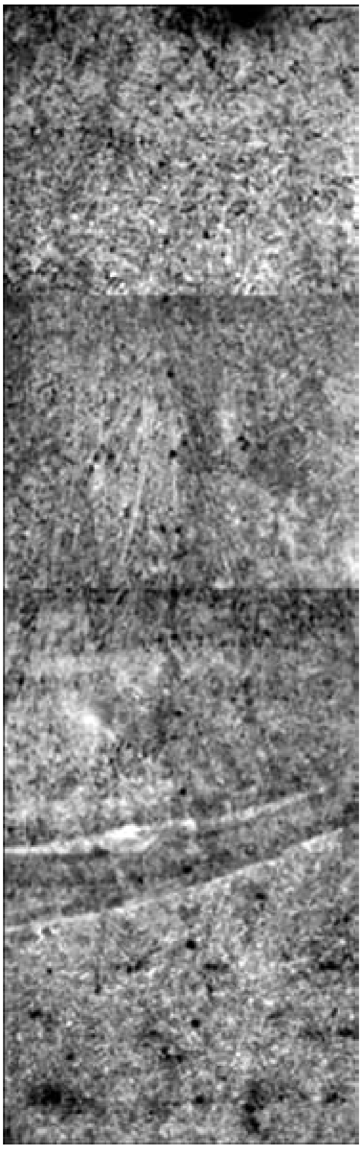

Figure 1. Change of microstructure (×50) of Fe–Cr–Mn steel in single-layer deposited state

where Δm_r, Δm_s is the weight loss of reference sample and deposited metal sample during the same wearing period, respectively.

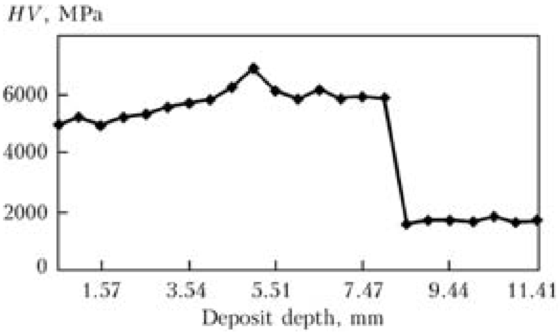

Figure 2. Distribution of microhardness by depth of single-layer deposited Fe–Cr–Mn metal

Steel 45 of hardness *HB* 180–190 was used as the reference sample. Testing for impact-abrasive wear was conducted in a unit shown in [10] in the environment of cast iron shot (0.5–1.5 mm particle size) with the speed of sample rotation of 2800 min^{-1}. Testing for abrasive wear was conducted by Brinell–Howard schematic. Relative impact-abrasive $\varepsilon_{im.a}$ and abrasive ε_a wear resistance was also determined by the above formula.

At flux-cored wire hardfacing the required formation of the deposited layer without any visible defects and good separability of the slag crust were ensured. Height of the deposited layer at one-layer hardfacing was 6–8, at two-layer — 11–15 and at three-layer 16–18 mm, respectively. Composition of the deposited metal, depending on the number of layers and welding-technological parameters, corresponds to the composition of steel of 20Kh(7–12)G(5–9)SF type. In the case of one- and two-layer hardfacing chromium content in the deposited metal was in the range of 6.41–7.98 %, that of manganese — 4.86–5.60 %.

Microstructure* of one-layer deposited metal consists of martensite and 15–35 % residual austenite A_{res} (see Figure 1). Structure is non-uniform across deposited metal thickness. Martensite has a pack (lath) structure, which is indicative of the fact that this is low-carbon martensite. Upper zone of small dimensions (0.6–0.8 mm) in one-layer deposit has fine-crystalline structure (see Figure 1). This is followed by a rather extended (2–5 mm) zone with clear-cut elongated predominantly towards the hardfaced surface crystals of about 0.02 mm thickness and about 0.3 mm length. The crystals are elongated in the direction opposite to that of heat removal in-depth of the base metal. Located under it is a uniform transition zone of 0.14 to 0.30 mm thickness that borders on the fusion zone, which is adjacent to the HAZ and further on is the ferritic-pearlitic structure of base metal.

Microhardness variation in-depth of deposited Fe–Cr–Mn steel is given in Figure 2. Higher values of microhardness correspond predominantly to martensite phase, and lower values — to austenite phase. Microhardness is equal to *HV* 5000–5600 MPa by deposit depth of 0.59–3 mm that corresponds to austenitic-martensitic structure with prevalence of the austenite component, whereas higher values of *HV* 5800–7000 MPa (at the depth of 3.5–8.5 mm) — to martensitic-austenitic structure with prevalence of martensite. This is due to some possible gradient of chromium and manganese concentration by the depth of the deposited layer, afftecting the position of martensite points (M_s and M_f), which predetermines the phase relationship between martensite and austenite. Then a transition zone from deposited metal to base metal is observed, characterized by an abrupt lowering of microhardness from *HV* 5900 MPa to *HV* 1600–1900 MPa and of hardness from *HRC* 46–48 to *HRC* 28, respectively.

---

*O.I. Trofimets participated in investigations.

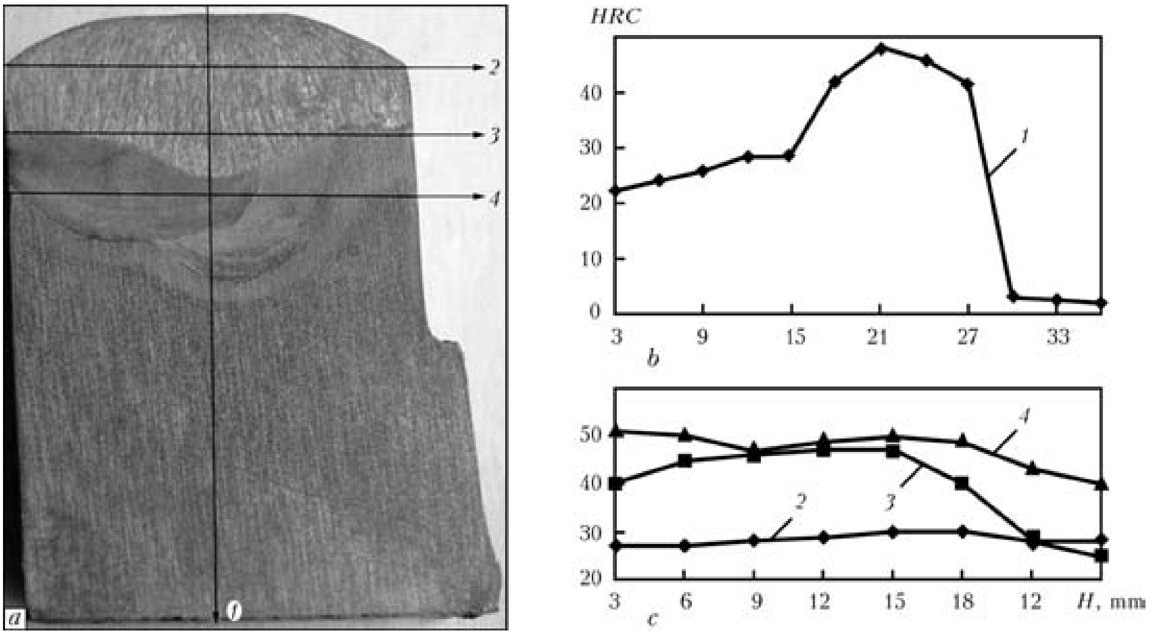

Figure 3. Macrostructure (*a*) and variation of hardness of transverse macrosection of multilayer deposited metal of 20Kh12G9SF type in vertical (*b*) and horizontal (*c*) directions (*1–4*)

Macrostructure of multilayer deposited metal* is given in Figure 3, where individual layers of deposited Fe–Cr–Mn metal are characterized by different degrees of etchability. The first (lower) layers have higher etchability compared to upper layers. This is attributable to different degree of their alloying: lower layers are less alloyed, as mixing and greater dilution of the weld pool by unalloyed steel of the base occurred in them. Macrostructure of the first (lower) layers features fine grains that may be due to the processes of metal recrystallization at heating, due to the heat of the deposited next layer. HAZ metal is of a lighter colour that is also attributable to recrystallization of the base metal under the fusion zone. The upper layer is characterized by a dendritic structure.

The panorama of the change of deposited Fe–Cr–Mn metal microstructure at multilayer hardfacing is given in Figure 4. Upper layer microstructure consists predominantly of austenite with carbide particle inclusions of $(Cr, Fe)_{23}C_6$, VC composition. Located under it are deposited metal layers with austenitic-martenisic and further on with martensitic-austenitic structure, also reinforced by carbide inclusions. In the upper deposited layers austenitic dendrites elongated normal to the surface are observed, which were growing in the direction opposite to that of heat removal. Austenite grains contain sliding lines and twins, as well as indications of ε-martensite structure, that is indicative of its deformation metastability, i.e. ability of self-strengthening at the expense of dynamic twinning and $\gamma \rightarrow \alpha'$ DMTT. Deposited metal composition is strongly influenced by the degree of penetration and fraction of base metal in the deposit metal. At multilayer hardfacing the composition of each layer is different, depending on the fraction of base metal involved in formation of the next deposited layer.

Change of the fraction of base metal (St3) in the deposited metal influenced the composition of the deposited layers by the content of chromium (8–12 %) and manganese (5.6–9.0 %), while carbon, silicon and vanadium content remained practically constant. Differences in alloying element content within the above limits influenced the position of martensite points (M_s, M_f) and phase composition, namely quenching martensite and metastable austenite content.

The predominantly austenitic structure of the deposit upper layer is indicative of the highest content of chromium and manganese in it, lowering martensite points M_s and M_f (apparently, below room temperature). Medium deposited layers have austenitic-martensitic structure, which is due, probably, to somewhat lower content of chromium and manganese compared to upper layers. Formation of martensitic-austenitic structure in the first deposited layers is indicative of an even lower content of alloying elements due to greater mixing with the base metal. As a result, martensite point M_s is above room temperature that leads to formation of predominantly quenching martensite at cooling with a small quantity of A_{res}. Thus, hardfacing parameters responsible for different penetration, fraction of base metal participation in weld pool formation, as well as number of deposit layers allow effectively controlling its chemical and phase composition. In its turn, the degree of deformation metastability of austenite and kinetics of $\gamma \rightarrow \alpha'$ DMTT [11] essentially depend on the ratio of martensite and austenite that determines formation of mechanical characteristics, and at optimum parameters an increased level of deposited steel wear resistance is achieved.

---

*N.E. Karavaeva participated in investigations.

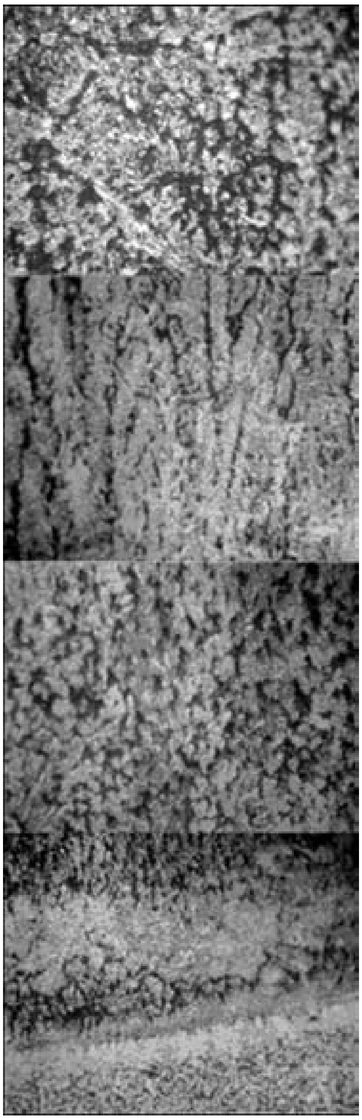

Figure 4. Change of microstructure (×170) of a transverse section of multilayer deposited Fe–Cr–Mn steel

Variation of hardness of transverse sections of deposited Fe–Cr–Mn steel in multilayer hardfacing is indirect confirmation of the nature of variation of phase composition and microstructure. Measurements were performed in keeping with the schematic, given in Figure 3, *a*, and change of hardness values in the vertical *1* and horizontal *2–4* directions is shown in Figure 3, *b*, *c*. Upper layer of the deposited metal is characterized by low hardness from *HRC* 22 up to *HRC* 28 that corresponds to the austenitic structure. Hardness increases by the layer depth. The highest hardness is found in the medium and lower layers of the deposited metal (*HRC* 42–47) (see Figure 3, *b*), having martensitic-austenitic structure. An abrupt lowering of hardness from approximately *HRC* 47 to *HRC* 5 takes place in the fusion zone, while martensitic-austenitic structure transforms into the initial ferritic-pearlitic structure.

Hardness variation corresponding to the considered layers is also observed in the horizontal direction (see Figure 3, *c*). Upper layer is characterized by the lowest hardness of *HRC* 28–30 with its uniform distribution, and medium and upper layers of the deposited metal have a higher hardness. Hardness of the medium and lower layers corresponds to *HRC* 46–48 and *HRC* 48–50. Hardness lowering to the left and right of the center along a horizontal shows an actual transition into the upper layer.

Mechanical properties of deposited steel

Type of deposited metal alloying	Hardness *HRC*	Impact toughness, KCU, MJ/m^2	ε	$\varepsilon_{im.a}$	ε_a
20Kh8G6ASF	34	0.35	1.22	3.44	1.4
20Kh12G9SF	33	0.42	1.05	3.52	1.5
08Kh20N10G7ST	19	1.10	1.10	2.10	0.9

Comparative testing of wear resistance of developed Fe–Cr–Mn and Fe–Cr–Ni deposited metal of austenitic class, as well as that deposited with imported wire of Sv-08Kh20N10G7ST type (applied for hardfacing the rolls of Pilger rolling mill at «Ilyich Metallurgical Works») at different wearing conditions was conducted. Mechanical properties of hardfaced steels (without heat treatment) are given in the Table. It follows from the Table that impact toughness of Fe–Cr–Ni deposited metal is 2 times higher than that of Fe–Cr–Mn that is due to purely austenitic stable structure with a low hardness (*HRC* 19), lower content of carbon and positive influence of nickel on the ductility and toughness properties [12].

Under the conditions of dry metal-to-metal friction (with heating of friction surface) wear resistance ε of deposited Fe–Cr–Mn metal is by 10–15 % higher than that of Cr–Ni metal, and under the conditions of abrasive and impact-abrasive wear $\varepsilon_{im.a}$ is 1.5–1.8 times higher. This is attributable to metastability of austenite in the structure of 20Kh8G6ASF steel and considerable self-strengthening of the surface layer during wearing due to running of $\gamma \to \alpha'$ DMTT. If before wearing deposited 20Kh8G6ASF metal contained 15 % of quenching martensite and 85 % of metastable austenite, after abrasive-impact wearing martensite content in the surface layer increased up to 41 %. Deposited 20Kh12G9SF metal contained 100 % of metastable austenite, and after wearing at dry friction 27.5 % of deformation martensite was detected in the surface layer of samples, alongside austenite. Obtained data are indicative of the fact that a higher level of service durability of the developed nickel-free deposited metal can be achieved, compared to Cr–Ni one, deposited with 08Kh20N10G7ST wire (deficit and much more expensive — approximately 1300–1500 USD/t).

Composition of deposited 20Kh12G9SF metal should be regarded as optimum in terms of complete implementation of the capabilities of $\gamma \to \alpha'$ DMTT for increase of wear resistance. Methods of heat treatment, chemico-thermal treatment, plasma or electron beam impact can be used to additionally create new heterophase-structural modifications to control the properties of the developed deposited metal.

CONCLUSIONS

1. Composition and structure of the deposited metal change by a certain law across the deposit thickness, depending on the number of deposited layers, degree of penetration and fraction of base metal in the deposit.

2. Modes of electric arc surfacing allows effective control of phase composition (austenite and martensite ratio), degree of metastability of the austenite component of deposited Fe–Cr–Mn metal, and, as a result, its mechanical properties.

3. Comparative testing showed an increased wear resistance of the deposited metastable Fe–Cr–Mn metal, compared to the known Cr–Ni composition of metal deposited using expensive and deficit 08Kh20N10G7ST wire.

1. Livshits, L.S., Grinberg, N.A., Kurkumelli, E.G. (1969) *Principles of alloying of deposited metal*. Moscow: Mashinostroenie.
2. Razikov, M.I., Tolstov, I.A., Kulishenko, B.A. (1966) Experience in application of deposited metal of 30Kh10G10T type for hardfacing of rapidly wearing parts. *Svarochn. Proizvodstvo*, **9**, 30–31.
3. Kulishenko, B.A., Shumyakov, V.A., Maslich, S.Yu. (1985) Application of martensite transformation in deformation for increase in wear resistance of deposited metal. In: *Hardfacing: experience and efficiency of application*. Kiev.
4. Filippov, M.A., Litvinov, V.S., Nemirovsky, Yu.P. (1988) *Steels with metastable austenite*. Moscow: Metallurgiya.
5. Kovalchuk, A.I., Oldakovsky, A.I., Malinov, L.S. et al. (1984) Increase in serviceability of cylinders of pilgrim mills by hardfacing using new flux-cored wire 35ZhN. *Svarochn. Proizvodstvo*, **7**, 12–13.
6. Malinov, L.S., Chejlyakh, A.P., Kharlanova, E.Ya. et al. (1995) Development and investigation of new flux-cored strip for hardfacing of bridge crane wheels. *Ibid*., **10**, 22–25.
7. Malinov, V.L., Chigarev, V.V. (1997) Influence of structure on wear resistance of deposited metal in different types of impact-abrasive wear. *Vestnik PGTU*, Issue 3, 141–144.
8. Cheiliakh, Y., Chigarev, V., Sheychenko, G. (2009) The creation of a new economical (nickel free) powder-like wire for surfacing made of metastable metal, self-strengthened during wearing. In: *Proc. of 1st Mediterranean Conf. on Heat Treatment and Surface Engineering in the Manufacturing of Metallic Engineering Components* (Sharm El-Sheikh, 1–3 Dec., 2009).
9. Akulov, A.I., Alyoshin, N.P., Chernyshov, G.G. (2004) *Welding. Cutting. Control*: Refer. Book. Vol. 2. Moscow: Mashinostroenie.
10. Chejlyakh, A.P., Olejnik, I.M. *Unit for testing on impact-abrasive wearing*. USSR author's cert. 1820300. Int. Cl. G01N 3/56. Publ. 07.06.93.
11. Chejlyakh, A.P. (2003) *Sparcelly alloyed metastable alloys and strengthening technologies*. Kharkov: KhFTI.
12. Meskin, V.S. (1964) *Principles of alloying of steel*. Moscow: Metallurgiya.