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REGULATION OF THE PHASE-STRUCTURAL COMPOSITION AND METASTABILITY OF AUSTENITE BY ALLOYING ELEMENTS AND PARAMETERS OF TEMPERING FOR INCREASE WEAR-RESISTANCE OF Fe-Cr-Mn DEPOSITED STEEL

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Abstract. It is shown that under the influence of alloying (Cr, Mn, C, Si, Ti, N, V) during electric arc hardfacing with a flux cored wire, as well as technological tempering parameters at temperatures of 300–700 °C, the phase-structural composition of the cost-saving alloyed deposited steel (from 0 to 75 % quenching martensite and austenite), the degree of metastability of austenite are regulated. At optimal parameters of alloying and tempering, an increased wear resistance is achieved due to the development of the deformation induced martensite $\gamma \rightarrow \alpha'$ transformation of austenite during the wear process, which causes the effect of self-strengthening during testing and operation. This is an important advantage of the developed metastable deposited steel before the deformation-stable metal.

Keywords: deposited steel, alloying, metastable austenite, martensite, tempering, wear resistance.

Introduction

In order to restore shapes and exploitation properties of machine parts and tools surface deposition is widely used, it being a process of deposition of a layer of metal or alloy upon a part’s surface by means of melting welding [1]. Electric-arc hardfacing with powder electrodes is especially efficient, as it has a series of advantages over other methods, including simplicity of the process, wide possibilities of obtaining various compositions of chemical composition and consequently properties and also high quality of deposited layers [2], [3]. Obtaining of deposited steel, surpassing in quality the part’s material seems to be the most promising approach to development of new hardfacing materials, it can be reached, primarily, by a correct selection of the content of alloying components. Then, alloying is to be efficiently done with inexpensive and wide-spread alloying components, like chromium, manganese, titanium, silicon and the like [2], [4]–[8]. Therefore, the task of cost-saving alloyed of deposited metal and simultaneous enhancement of operational properties is an actual scientific and practical task of modern materials science.

Problem Statement

Among a wide range of known surfacing materials, it is necessary to distinguish those that provide a hardfacing metastable deposited metal on the renewal parts, which has a number of important advantages.

Metastable deposited metal usually possesses metastable austenite structure [4], or its combination with other structures, especially hard phases and structures (martensite, bainite, carbides, carbide-nitrides, borides etc.) [9].

Metastability of deposited metal is revealed in realization of deformation-induced martensite $\gamma \rightarrow \alpha'$ transformation at wear (DIMTW) at testing or service (under deforming action of the wear medium) and also at cold-working of austenite and martensite. It causes the effect of strong self-strengthening of the surface working layer directly at the process of testing or operation and is an important advantage of metastable deposited metal over deformation-stable materials. Deformation martensite, generated from metastable austenite because of DIMTW differs from quenching martensite by its better disperse, density of dislocations, the value of micro-distortions and as a result it has an increased wear resistance. Relaxation of micro-stresses goes on simultaneously with the process of realization of DIMTW, it promoting better reliability and workability of such deposited metal [9], [10].

Review of Modern Information Sources on the Subject of the Paper

To date, many works have been devoted to the development of powder surfacing materials of different chemical composition and to the investigation of deposited metal with a metastable structure [2–4], [9], [10]. However, the possibilities of cost-saving alloying and methods of thermal and other types of treatment of deposited metal as additional sources of hardening remain insufficiently investigated and far from exhausted. In this there are large reserves, the disclosure of which will significantly improve the mechanical and operational properties with cost-saving alloying of wear-resistant deposited steels.

Objectives and Problems of Research

The objective of the work is to research the possibilities of regulating the phase-structure composition and the degree of metastability of the austenite component by alloying and parameters of tempering to improve the mechanical and operational properties of the deposited cost-saving alloyed Fe-Cr-Mn steels.

Materials and methods of research

To hardfacing wear resistant DM with a metastable structure of different composition, the flux-cored wires of the grades PP-Hp-20Cr10MnSiTiNV, PP-Hp-20Cr12Mn10SiTiNV and PP-Hp-12Cr13Mn12SiTiNV were developed [8], [11]. With their use, electric arc hardfacing (surfacing) of Fe-Cr-Mn steels of various composition and structural classes: martensitic-austenitic, austenitic-martensitic and austenitic, containing alloying elements within the following ranges (in mass %): 0.12–0.2 C; 8–13 Cr; 6–12 Mn; with small additions of Si, Ti, N, V; content sulfur ≤0.01; phosphorus ≤0,03. Due to alloying and variation in the chemical composition, the phase composition and the degree of deformation metastability of the austenite phase and as a result – mechanical and operational properties of deposited steel were regulated [12–14].

Electric arc hardfacing was carried out by the aforementioned powder wires with a diameter of 4 mm at the A1401 welding machine with a VDU 1001 power supply with reverse polarity. To protect the weld metal, fluxes of AN-348 and AN-60 grades were used. The hardfacing modes are as follows: welding current 320–480 A, arc voltage 28–32 V, the deposition rate was V_{dep} = 22–32 m/h, the number of layers 3–4. From deposited steels samples for research and testing were made.

Analysis of the chemical composition was carried out using the “Spectrovac 1000” and “SpectroMAXx” vacuum quantometers using the spectral method. The phase deposited steels analysis was performed on a DRON-3 X-ray diffractometer in iron K_α-radiation, in the angular interval $2\theta = 54–58$ deg. Macroscopic analysis was carried out on transverse and longitudinal macro-sections of the deposited steels with deep etching with a mixture of acid solutions (100 ml HCl, 10 ml HNO_3 and 100 ml H_2O). The microstructure was researched on metallographic microscopes MMR-2 and “Neophot-21” at enlargement from 50 to 500 times. The microhardness of the structural components was measured on a PMT-3 microhardness gauge by pressing a diamond tetrahedral pyramid under a load of 1.96 N, and the hardness of deposited steels on a Rockwell hardness gauge with a load of 1500 N (HRC) and 600 N (HRA). Dynamic bending tests were carried out on a pendulum copre IO5003 on samples with a measuring of 10×10×55 mm with a U-shaped notch.

Tests for wear were carried out under various friction and wear conditions. In the case of dry friction metal on metal, they were carried out on a MI-1M machine on samples of deposited steels measuring 10×10×27 mm in accordance with the scheme test sample – a roller (control body) rotating at a speed of 500 min^{-1} (linear speed in the friction zone – 1.31 m/s, the friction path is 1965 m). Wear time was:

between two weightings – 5 minutes, the total – 25 minutes. Weighting was performed with an error of up to 0.0001 g. Relative wear resistance was determined by the formula:

$$e = \frac{\Delta m_{st}}{\Delta m_{sample}}, \tag{1}$$

where Δm_{st}, Δm_{sample} is the loss of mass, respectively, of the standard sample and the sample of the deposited steels for the same wear time. As a standard sample, steel 45 (contain 0.45 % carbon) was used with a hardness of HB180–190. Tests for impact abrasive wear were carried out on a facility described in [15] in a medium of cast iron shot (fractions 0.5–1.5 mm) at a samples rotation speed of 2800 min^{-1}. Tests for abrasive wear were carried out according to the Brinell-Howarth scheme in quartz sand. The relative impact-abrasive wear resistance ($\varepsilon_{i.-a.}$) and abrasive wear resistance (ε_a) was also determined by the formula (1).

Main Material Presentation

Effect of alloying wear resistant deposited steel. For different wear conditions (dry wear of metal against metal, abrasive and impact-abrasive wear) economically-alloyed compositions of flux-cored wire, ensuring obtaining of metastable Fe-Cr-Mn deposited steels with prescribed relation of austenite and martensite were developed at PSTU.

The chemical composition of deposited steels is obtained by means of carbon content (0.1–0.45 mass. %) and alloying elements (firstly, chromium 8–14 mass. %, and manganese 6…13 mass. %), upon that the degree of metastability of the austenite component of structure depends, it determining the kinetics of $\gamma \rightarrow \alpha'$ DIMTW and formation of wear-resistance. The phase composition (quantity of quenching martensite (*M*) and metastable austenite (*A*) of deposited steels is shown in Table 1.

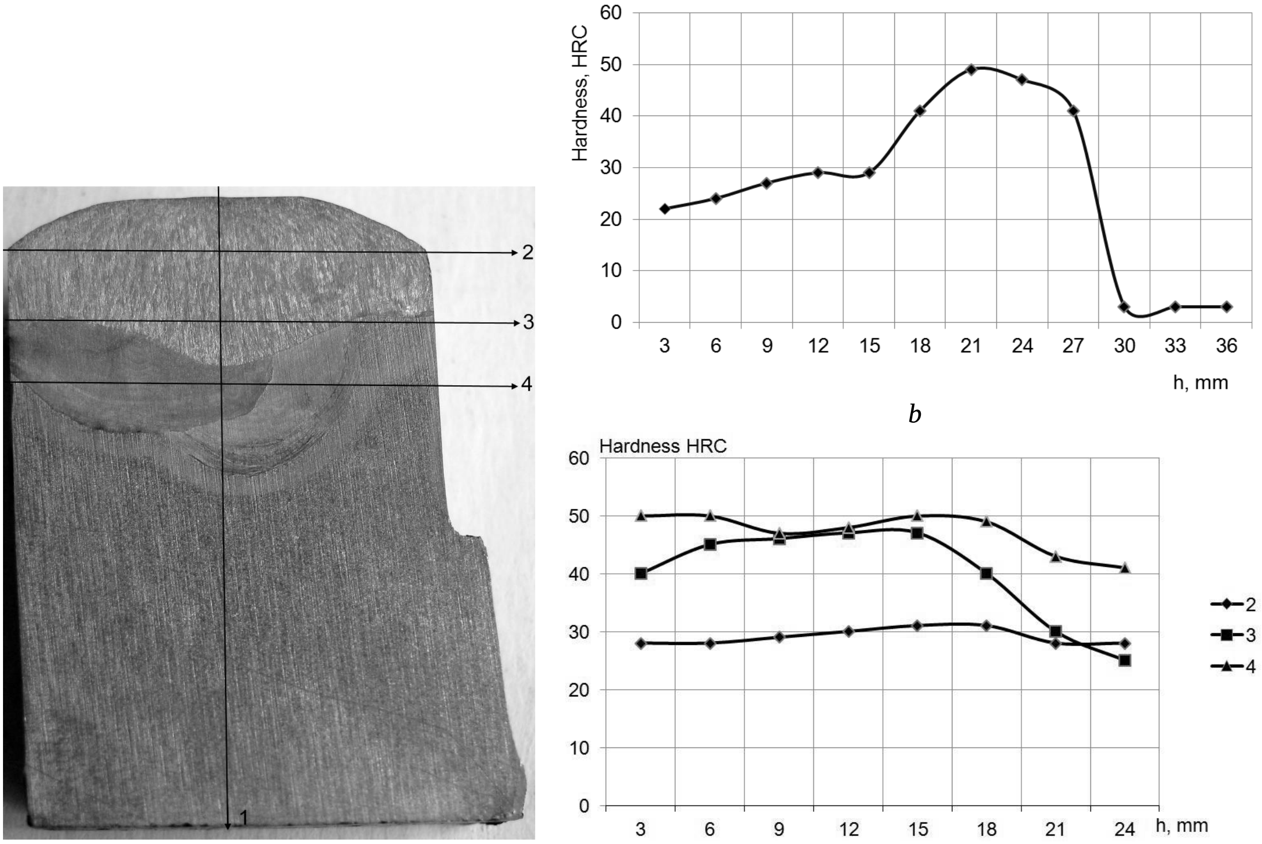
Table 1

Phase composition of Fe-Cr-Mn of deposited steel

Steel grade	Number of phases, %	
	M	*A*
20Cr8Mn6SiTiNV	75	25
20Cr10Mn8SiTiNV	20	80
20Cr12Mn9SiTiNV	10	90
12Cr13Mn12SiTiNV	0	100
45Cr13Mn12SiTiNV	0	100

Macrostructure of multilayer deposited steels grade 20Cr12Mn9SiTiNV is given in Fig. 1, where individual layers of it are characterized by different degrees of etching ability. The first (lower) layers have higher etching ability 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. Heat affected zone (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 steel 20Cr12Mn9SiTiNV microstructure at multilayer hardfacing is given in Fig. 2. Upper layer microstructure consists predominantly of austenite with carbide particle inclusions of $(Cr, Fe)_{23}C_6$, VC composition. Located under it are deposited steels 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 $\gamma \rightarrow \alpha¢$ DIMTW.



a *c*

***Fig. 1.** Macrostructure (a) and variation of hardness of transverse macrosection of multilayer deposited steel grade 20Crl2Mn9SiTiNV in vertical (b, direction 1) and horizontal (c, directions 2, 3, 4)*

Deposited steel composition is strongly influenced by the degree of penetration and fraction of base steel in the deposited steel. 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.

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_{ret}.

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 Fig. 1, *a*, and change of hardness values in the vertical and horizontal 2–4 directions is shown in Fig. 1, *b*, *c*. Upper layer of the deposited metal is characterized by low hardness from HRC22 up to HRC28 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 steel (HRC42–47) (see Fig. 1, *b*), having martensitic-austenitic structure. An abrupt lowering of hardness from approximately HRC47 to ~HRC5 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 Fig. 1, *c*). Upper layer is characterized by the lowest hardness of HRC28–30 with its uniform distribution, and medium and upper layers of the deposited steels have a higher hardness. Hardness of the medium and lower layers corresponds to HRC46–48 and HRC48–50. Hardness lowering to the left and right of the center along a horizontal shows an actual transition into the upper layer.

It can be seen on the Table 1 that under the influence of alloying elements and carbon it is possible to modify the obtained phase composition of Fe-Cr-Mn deposited steels within a wide range (from 25 to 100 % of metastable austenite). Judging by service conditions it helps to control the kinetics and the degree of transformation of metastable austenite into deformation martensite, due to $\gamma\rightarrow\alpha'$ DIMTW. Deposited steels of 12Cr13Mn12SiTiNV grade is characterized mostly by austenite microstructure of the upper layer (small amount of quench martensite is possible) (Fig. 3).

Then along its depth column austenite dendrites are observed, a line of melting with the base metal is clearly seen (Steel St.3 content ~0.2 % C). Alongside with reduction of chromium and manganese content in deposited steels the amount of quenching martensite is increased up to ~45 % (see Table 1). The mechanical properties of Fe-Cr-Mn deposited steels (without heat treatment) are summarized in Table 2. Relative wear-resistance under conditions of dry friction sliding of metal against metal (ε) (with heating of the frictional surface), abrasive (ε_a) and impact-abrasive ($\varepsilon_{i\text{-}a}$) wear resistance of deposited steels 20Cr12Mn9SiTiNV is higher by 20–25 % than that of austenitic-martensitic steel 20Cr8Mn6SiTiNV (Table 2). It is explained by optimal metastability of austenite and sufficient self-strengthening of the surface in the process of wear, owing to immense volume of $\gamma\rightarrow\alpha'$ DIMTW with formation of ~36 % of deformation martensite.

In deposited steel of 12Cr13Mn12SiTiNV grade ε and $\varepsilon_{i\text{-}a}$ are slightly lower, due to lesser carbon content and small growth of deformation martensite ΔM = 16 % and $\Delta M_{i\text{-}a}$ = 26 %, accordingly. The highest values of relative wear-resistance (ε and $\varepsilon_{i\text{-}a}$) are reached for deposited steel 45Cr13Mn12SiTiNV grade because of increased carbon content and bigger volume of $\gamma\rightarrow\alpha'$ DIMTW realization in the surface layer: quantity of deformation martensite ΔM = 27 % *and* $\Delta M_{i\text{-}a}$ = 39 % (see Table 2).

Comparative testing of wear resistance of developed Fe-Cr-Mn and Fe-Cr-Ni deposited steel of austenitic class, as well as that deposited with imported wire of Sv-08Cr20Ni10Mn7SiTi type at different wearing conditions was conducted [12]. Mechanical properties of hard-faced steels (without heat treatment) are given in the Table 3.

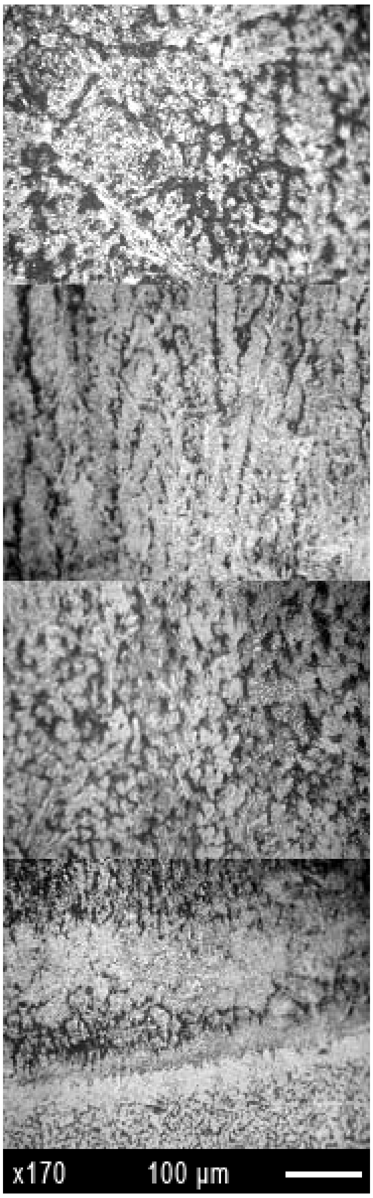


Fig. 2. *Change of microstructure of a transverse section of multilayer deposited steel 20Cr12Mn9SiTiNV*

It follows from this Table 3 that impact toughness of Fe-Cr-Ni deposited steel 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.

Under the conditions of dry metal-to-metal friction sliding (with heating of friction surface) wear resistance Ε of deposited steel Fe-Cr-Mn is by 10–15 % higher than that of Cr-Ni metal, and under the conditions of abrasive and impact-abrasive wear $\varepsilon_{i\text{-}a}$ is 1.5–1.8 times higher. This is attributable to metastability of austenite in the structure of 20Cr8Mn6SiV steel and considerable self-strengthening of the surface layer during wearing due to running of $\gamma\rightarrow\alpha'$ DMTW. If before wearing deposited steel grade

20Cr8Mn6SiV 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 steel 20Cr12Mn9SiV contained 100 % of metastable austenite, and after wearing at dry friction sliding 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 steel can be achieved, compared to chromium-nickel one, deposited with 08Cr20Ni10Mn7SiTi wire (deficit and much more expensive — approximately 1300–1500 USD per ton).

Table 2

Hardness, wear-resistance and changes in phase composition at wear of deposited Fe-Cr-Mn steel

Grade of deposited steel	HRC	A, %	ε	ΔM, %	e_{i-a}	ΔM_{i-a}, %
20Cr8Mn6SiTiNV	34	75–80	2.0	10	5.1	21
20Cr12Mn9SiTiNV	33	95	2.53	20	5.5	36
12Cr13Mn12SiTiNV	24	100	2.4	16	4.8	26
45Cr13Mn12SiTiNV	26	100	6.4	27	14.2	39

Table 3

Mechanical properties of deposited steel

Grade of deposited steel	Hardness *HRC*	Impact toughness, KCU, MJ/m^2	ε	ε_{i-a}	ε_a
20Cr8Mn6SiV	34	0.35	1.22	3.44	1.4
20Cr12Mn9SiV	33	0.42	1.05	3.52	1.5
08Cr20Ni10Mn7SiTi	19	1.10	1.10	2.10	0.9

Despite the fact that electric-arc hardface of metastable deposited steel with rational compositions generally is capable of solving the problems of parts restoration and simultaneous improvement of wear-resistance there exist some possibilities of additional surface strengthening by means of various methods of thermal and other actions [16, 17].

Effect of tempering of the wear resistant deposited metal. High-temperature tempering of deposited steel is the most wide-spread method of heat treatment. It is performed mainly for relieving of internal deposition stresses and decomposition of quenching structures (martensite, bainite), obtained at cooling of alloyed deposited steel for ferrite-carbide structure, in case their formation is not desirable. However, tempering not only makes it possible to relieve stresses, but also regulate the degree of metastability of the austenite component by modifying its temperature within 200–700 °C range and holding time from 0.5 to 6 hours, thus modifying its ability to DIMTW. Tempering may be applied both after hardfacing (for deposited steel of austenite and austenite-carbide classes) and after preliminary quenching of deposited steel from increased temperatures for special obtaining of austenite metastable structure.

So, deposited steel tempering of 20Cr10Mn8SiTiNV grade of mostly austenite structure (and some amount of quenching martensite) at relatively low temperatures 200–350 °C increases the stability of austenite by decreasing the level of internal stresses, blocking of dislocations by admixtures atoms. On the contrary the processes of formation of carbide and carbon-nitride phases at tempering temperatures 400–700 °C, when ageing processes are developed cause depletion of austenite by carbon and alloying elements. It raises M_s point and destabilizes austenite with regard to martensite transformation at cooling and prior to DIMTW. As a result at cooling from tempering temperatures some amount of quenching martensite is formed, it being accompanied with increase in hardness of the surface layer from HRC30 to HRC38.

Impact toughness is steadily lowered at that from 0.42 up to 0.23 MJ/m^2 that is explained by embrittling action of precipitated carbides. The greater amount of precipitated carbides the greater is depletion of austenite by carbon and alloying elements and the bigger its metastability is, as well as its inclination to DIMTW and, hence, to self-strengthening.

The smallest inclination to DIMTW is observed for austenite deposited steel without heat treatment or after low-temperature tempering (200–300 °C), which is proved by relatively small amount of deformation martensite in the surface layer at wear (Table 4).

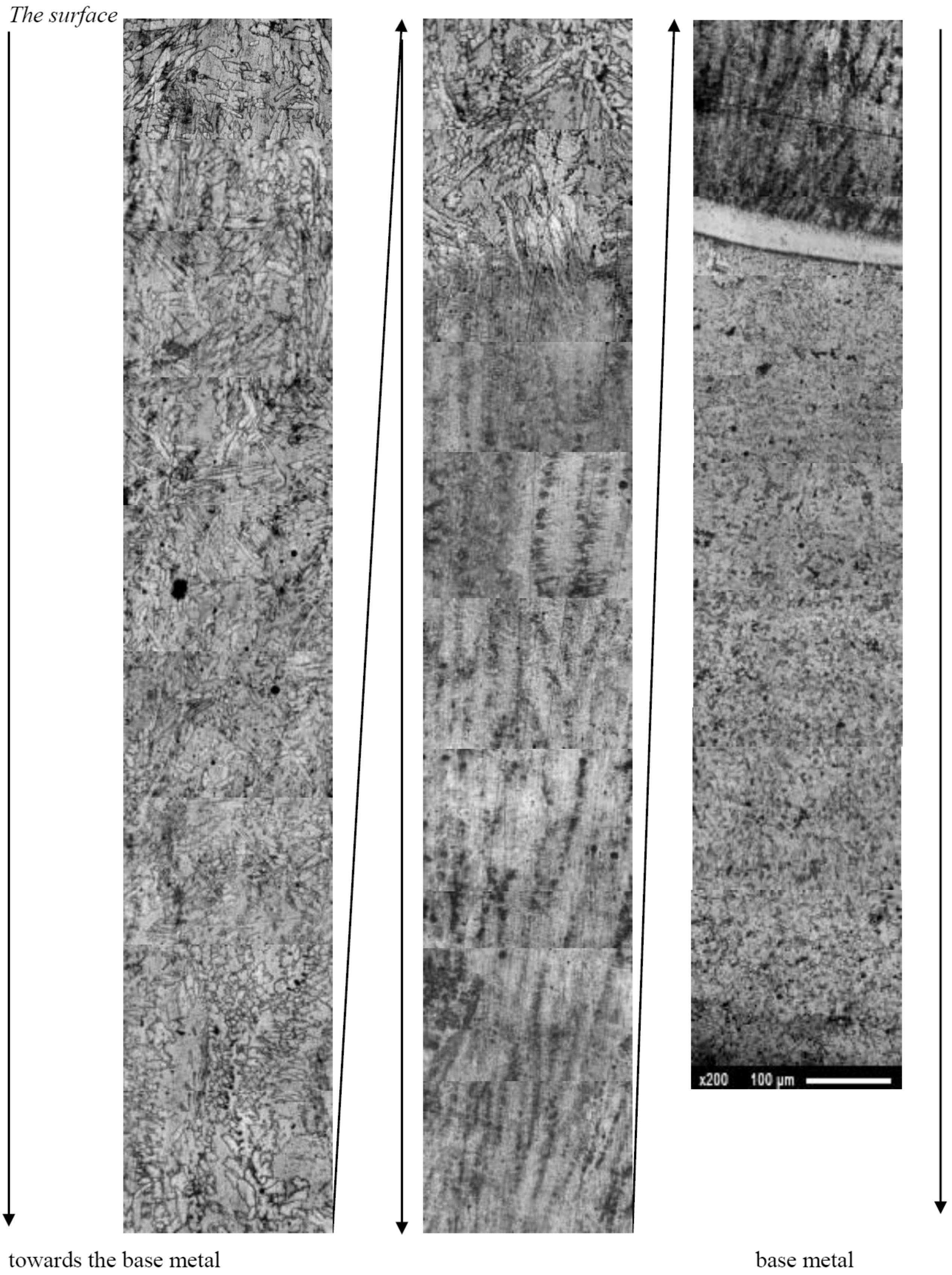


Fig. 3. *The microstructure of deposited steel 12Cr13Mn12SiTiNV*

It is worthwhile mentioning here that inclination of deposited steel to DIMTW is different under different wear conditions; it is explained by a difference in stress-strain state. The smallest volume of DIMTW is formed in case of sliding friction (at 69 N load) ~12 % of deformation martensite. The highest volume – 32 % of deformation martensite for the same deposited steel is formed at impact-abrasive action. The bigger tempering temperature is, hence, the bigger degree of austenite destabilization, the greater is growth of martensite amount inside the surface layer at different wear conditions: dry sliding friction, impact-abrasive and abrasive (ΔM, ΔM_{i-a}, ΔM_a) (see Table 4).

Table 4

Influence of tempering temperature upon growth of amount of deformation martensite at wear testing of deposited steel of 20Cr10Mn8SiTiNV at different conditions

$T_{temper.}$, ºC	Dry friction	Impact-abrasive	Abrasive
	ΔM, %	ΔM_{i-a}, %	ΔM_a, %
-	12	28	32
300	21	25	30
600	30	41	36

Complex testing of wear-resistance at different conditions proved that tempering at 300–500 ºC temperature range has little effect on abrasive and impact-abrasive wear resistance of deposited steel of 20Cr10Mn8SiTiNV grade, while at 600 ºC it substantially (1.2–1.6 times) raises it (Fig. 4), it being in well agreement with growth of amount of deformation martensite, formed in the surface (working) layer (see Table 4). Under conditions of dry sliding friction (with heating of the contact area with counter-figure) tempering from even 300 ºC increases wear-resistance of deposited steel ~ by 2 times approximately, as compared to the state without application of heat treatment, and at tempering at 500–600 ºC this value is 2.7–2.8 higher (see Fig. 4).

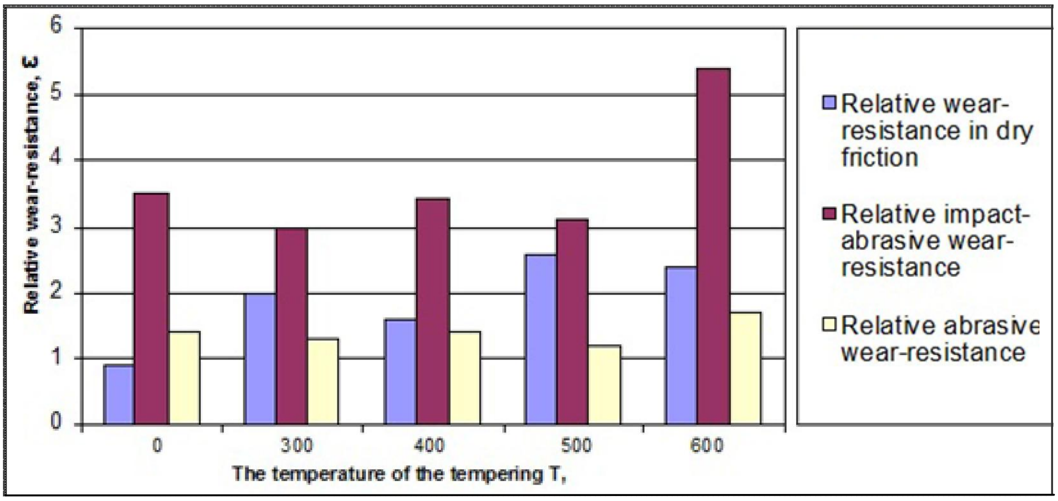


***Fig. 4.** Relative wear-resistance of deposited steel of 20Cr10Mn8SiTiNV grade at different types of wear after tempering at different temperatures*

It is explained by disperse strengthening due to precipitation of carbides (carbon-nitrides), destabilization of austenite, connected with some growth in the amount of quenching martensite and, what is most important, by activation of $\gamma \rightarrow \alpha'$ DIMTW inside the surface layers, leading to an increase in formation of deformation martensite. In order to obtain mostly austenitic structure of deposited steel (if it is not formed at deposition), of e.g. 20Cr8Mn6SiTiNV grade (75 % of martensite and 25 % A_{ret}), it is necessary to perform quenching from raised temperatures (for example at temperature 1100 °C holding

25 min. to increase the amount of austenite to 86 % and to recrystallize the dendritic structure) and then modify austenite metastability by tempering 300–700 °C.

The microstructure of the upper layers of the HM 20Cr8Mn6SiTiNV after quenching and tempering at 300–400 °C consists mainly of recrystallized austenite (86 % A + 14 % M) (Fig. 5), obtained as a result of $\alpha \rightarrow \gamma$ transformation, dissolution of carbides and carbonitrides in austenite upon heating under quenching. In the depth of the deposited steel, the structure gradually becomes austenitic - martensitic and closer to the melting zone – martensitic – austenitic. Therefore, the hardness of the deposited steel changes from HRC25...32 on the surface to HRC48–52 near the melting zone.

With an increase in tempering temperature from 500 to 700 °C, the martensite of quenching (~ 40 % M) appears in the microstructure of the surface layer of deposited steel (Fig. 5, *c*, *d*), owing to an increase martensite point M_s above room temperature, which is explained by depletion of austenite by carbon and alloying elements in bond with the precipitation of carbides and carbonitrides. As a result, the hardness of deposited steel increases from HRC28 to HRC40. At the same time, the relative wear resistance (ε_a, ε and $\varepsilon_{i\text{-}a}$) as a whole increases in comparison with the quenched state (Table 5).

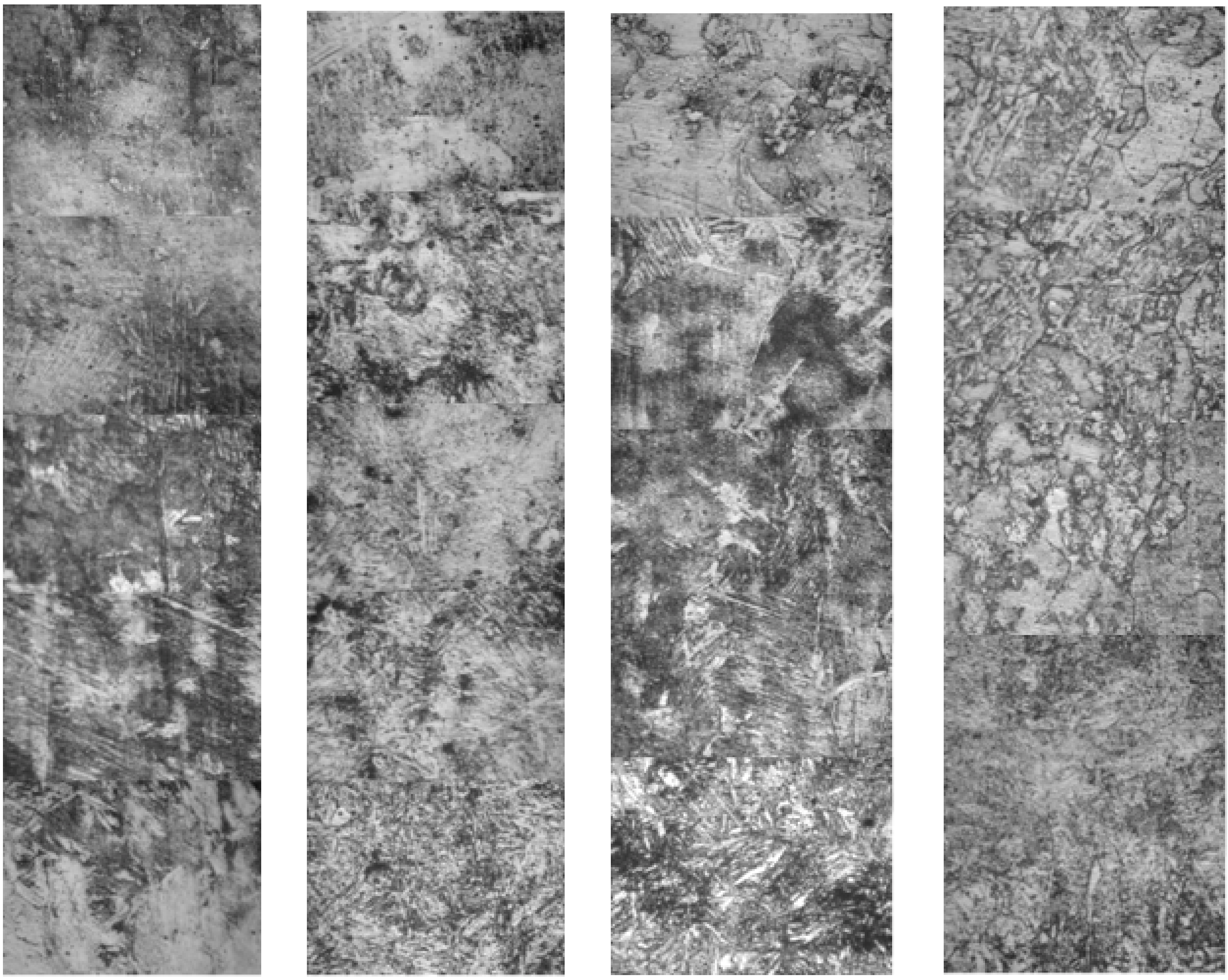

a *b* *c* *d*

Fig. 5. *Microstructure of the deposited steel 20Cr8Mn6SiTiNV after quenching 1100 ºC (a), and tempering at different temperatures: b – 400 ºC; c – 600 ºC; d – 700 ºC; x500*

The character of influence of tempering temperature, at that, upon development of transformation, described above and the obtained regularities of formation of wear-resistance are on the whole similar to the ones considered above. Optimal tempering modes at 600–700 ºC temperature after quenching allows activation of DIMTW austenite kinetics and drastic raising in wear-resistance of metastable deposited steel, e.g. for 20Cr8Mn6SiTiNV grade at different wear conditions (Table 5).

Table 5

Influence of tempering after quenching at 1100 ºC upon mechanical properties of deposited steel 20Cr8Mn6SiTiNV at different wear conditions

T_{temper}, ºC	Relative wear resistance			Hardness, HRC
	Dry sliding friction, ε	Abrasive, ε_a	Impact-abrasive, $\varepsilon_{i\text{-}a}$	
–	0.8	1.4	1.5	28
300	1.6	1.4	2.6	27
400	1.8	1.5	3.6	30
600	1.8	1.8	3.9	27
700	2.3	1.32	2.65	40

The effect of self-strengthening of such metastable deposited steel at different wear conditions is verified by an increase in growth of the amount of strain-induced martensite (ΔM, $\Delta M_{i\text{-}a}$, ΔM_a) with an increase in tempering temperature up to 600 ºC (Table 6) because of destabilization of austenite and γ→α' DIMTW is activated.

The results given in Table. 6 show that the highest abrasive wear resistance ($\varepsilon_a = 1.8$) and impact-abrasive wear resistance ($\varepsilon_{i\text{-}a} = 3.9$) corresponds to the largest increase in martensite deformation – $\Delta M_a = 36$ % and $\Delta M_{i\text{-}a} = 41$ %, respectively, in the surface layers in the contact zone. This confirms the significant contribution of DIMPTW to the formation of increased wear resistance of the deposited steel with a metastable structure.

Table 6

Influence of tempering temperature upon growth on martensite amount at deposited steel wear tests of 20Cr10Mn8SiTiNV at different wear conditions

$t_{temper.}$, ºC	Dry friction	Abrasive	Impact-abrasive
	ΔM, %	ΔM_a, %	$\Delta M_{i\text{-}a}$, %
–	12	32	28
300	21	30	25
600	30	36	41

Conclusions

1. Under the influence of alloying (mass. %): 8...14 Cr, 6...10 Mn, 0.18...0.25 C, 0.4...0.95 Si, small additions (N, Ti, V) form the microstructure of the wear resistant deposited Fe-Cr-Mn steel with a regulated amount of martensite and austenite in the range from 0 to 75 %, degree of austenite metastability, and its deformation induced martensite γ → α' transformation at wearing provides enhanced properties.

2. The chemical, phase compositions and the structure of the deposited steel vary regularly with respect to the overlaying section, depending on the number of deposited layers, the degree of penetration and the fraction of participation of the base steel (St.3) in the deposited steel, which is advisable to regulate the electric arc hardfacing (I, U, $V_{dep.}$) by flux cored wires, and effectively control the amount of austenite and martensite over a wide range (from 0 to 75 %), the degree of austenite metastability and, as a result, mechanical and operational properties.

3. It is shown that the technological parameters of tempering at 300–700 °C can effectively regulate the quantitative relationship between martensite and austenite in the deposited steel, which regulates the degree of austenite metastability and determines the formation of mechanical and operational properties. Under optimal regimes of treatment their enhanced characteristics are achieved.

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