

МАТЕРІАЛОЗНАВСТВО

УДК 621.78

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Toru Noguchi⁵, Abubaker Abulgasem Odenay⁶, Abdulmagid M.R. Kabellu⁷**SURFACE MODIFICATIONS FOR WEAR RESISTANT STEELS BY METHODS
OF INDUCED METASTABLE AUSTENITE STRUCTURES**

The problem of creation of new economical surfacing material, not containing expensive components (Ni, Mo, Nb, W and others) and new strengthening technologies for steels is quite actual and modern for many countries. One of the way of solving this problem is to develop metastable austenite conditions and realize them in the process of wear Deformation Induced Martensite $\gamma \rightarrow \alpha'$ Transformation (DIMIT), accompanied with emission of carbide, carbon-nitride or inter-metallic compound of expensive phases inside the surface layer. The purpose of this work is to develop new economical (nickel free) powder surfacing materials and new technologies of surface hardening providing strain self-hardening during the wearing process for increasing characteristic and wear resistant of the surfacing parts. New ways and processes of surface modifications of metastable-austenite phase-structural states were developed: electrode-arc surfacing with powder electrodes of metastable Fe-Cr-Mn steel grades of austenite, austenite-martensite or martensite-austenite classes without application of heat treatment, combined with subsequent quenching, ageing, thermo-cyclic treatment for regulating the degree of austenite metastability (depending upon the required wear resistance and other mechanical properties); plasma, electron-beam or laser treatments with different degree of surface melting (if necessary combined with volume heat treatment) of alloyed steels of different structural classes and designation.

Keywords: electrode-arc surface, plasma, electron-beam treatment, metastable austenite, martensite, wear resistant.

Я. Чейлях, О. Чейлях, В. Чигарьов, К. Шиміцу, Т. Ногучі, А. Оденая, А. Кабеллу. Поверхневі модифікації зносостійких сталей методами індукування метастабільних аустенітних структур. Проблема створення нових економічних наплавлювальних матеріалів, що не містять дорогі компоненти (Ni, Mo, Nb, W та інші) і нових зміцнюючих технологій для сталей актуальні і сучасні для багатьох країн. Одним з шляхів вирішення цієї проблеми - створювати метастабільні стани аустеніту і реалізовувати в процесі зношування деформаційні мартенситні $\gamma \rightarrow \alpha'$ перетворення (ДМП), що супроводяться виділенням карбідів, карбонітридів або інших інтерметаллідних з'єднань із збагачених фаз в наплавлених шарах. Мета цієї роботи - розвинути нові економічні (безнікелеві) порошкові наплавлювальні матеріали і нові технології поверхневого зміцнення, що забезпечують ефекти деформаційного самозміцнення в процесі зношування для підвищення характеристик і зносостійкості поверхневих шарів. Отримали розвиток нові шляхи і процеси створення поверхневих модифікацій фазово-структурних станів метастабільного аус-

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меніту: електро-дугове наплавлення порошковими електродами метастабільних Fe-Cr-Mn марок сталей аустенітного, аустенітно-мартенситного і мартенситного класів, без термічної обробки, і у поєднанні з подальшими гартуванням, старінням, термоциклічною обробкою для регулювання ступеня метастабільної аустеніту (залежного від необхідного опору зношуванню та інших механічних властивостей); плазмова, електронно-променева або лазерна обробка з різним ступенем нагріву і плавлення (при необхідності у поєднанні з регулюючою термічною обробкою) легованих сталей різних структурних класів і призначення.

Ключові слова: електро-дугове наплавлення, плазмова, електронно-променева обробка, метастабільний аустеніт, мартенсит, зносостійкість.

Я. Чейлях, А. Чейлях, В. Чигарев, К. Шимитцу, Т. Нозучи, А. Оденау, А. Кабеллу. *Поверхностные модификации износостойких сталей методами индуцирования метастабильных аустенитных структур.* Проблема создания новых экономичных наплавочных материалов, не содержащих дорогие компоненты (Ni, Mo, Nb, W и другие) и новых упрочняющих технологий для сталей актуальны и современны для многих стран. Одним из путей решения этой проблемы - создавать метастабильные состояния аустенита и реализовывать в процессе изнашивания деформационные мартенситные $\gamma \rightarrow \alpha'$ превращения (ДМП), сопровождающиеся выделением карбидов, карбонитридов или других интерметаллидных соединений из обогащенных фаз в наплавленных слоях. Цель этой работы - развивать новые экономичные (безникелевые) порошковые наплавочные материалы и новые технологии поверхностного упрочнения, обеспечивающие эффекты деформационного самоупрочнения в процессе изнашивания для повышения характеристик и износостойкости поверхностных слоев. Получили развитие новые пути и процессы создания поверхностных модификаций фазово-структурных состояний метастабильного аустенита: электро-дуговая наплавка порошковыми электродами метастабильных Fe-Cr-Mn марок сталей аустенитного, аустенитно-мартенситного и мартенситного классов, без термической обработки, и в сочетании с последующими закалкой, старением, термоциклической обработкой для регулирования степени метастабильности аустенита (зависящего от требуемого сопротивления изнашиванию и других механических свойств); плазменная, электронно-лучевая или лазерная обработки с различной степенью нагрева и плавления (при необходимости в сочетании с регулирующей термической обработкой) легированных сталей различных структурных классов и назначения.

Ключевые слова: электро-дуговая наплавка, плазменная, электронно-лучевая обработка, метастабильный аустенит, мартенсит, износостойкость.

Raising of problem. The problems of saving expensive and scarce alloying components, such as nickel, molybdenum, vanadium, cobalt, etc. when producing steels and alloys and improving their physical-mechanical and operational properties present great scientific and practical interest.

Increasing of the wear resistance and service lives of the various fast-wearing parts, operating parts of dividing mechanisms, excavating and tillage machines, rolls of the rolling mills and the other metallurgical equipment, - is quite actual scientific and production problem. Application of the electrode-arc surfacing - is one of the most efficient ways of the recovering worn-out parts.

As the most wide-spread wear resistance surfacing materials are used high-alloys and powder electrodes a great number of which contains expensive and deficit in many countries alloying components - nickel, molybdenum, vanadium, tungsten, niobium, cobalt and others [1] such saving is not always justified.

Analysis of the last researches and publications. Using the principle of metastable austenite, which was grounded by I. Bogachev, R. Mints [2], it is a very perspective way of developing special steels with metastable austenite in the researches of the works [3-6]. It is possible to apply the principle of metastable austenite not only to limited grades of special steels but also for other functional alloys, such as surfacing metastable steels, tool steels, alloyed cast-irons. For example, in high manganese erosive-resistant cast iron SCI-VMn with spherical carbides the work-hardening effects on the

material surface happened – reaching 804 HV from the initial hardness of 533 HV owing to austenite – martensite strain-induced transformation [7].

Application of the new generation alloys solves not only the problem of saving expensive and scarce alloying components (Ni, Mo, Nb, W and others) but also improves operational properties of industrial equipment and machine parts. The alloys with metastable structure possess self-organizing capability when loaded during operation. Programmed kinetics of deformation-induced phase transformations (DIPT: DIMT, deformation induced dynamic aging (DIDA)) in the alloys and surface strengthened layer under the influence of operational conditions allows to achieve a unique combination of extremely high mechanical and operational properties, resulting in efficient alloying and even complete exclusion of expensive components from their composition [8].

Materials and methods of research. The new economical powder-like (nickel-free) wire for surface deposition of wear-resistant Fe-Cr-Mn steels was created and the appropriate steels and tool steel 150Cr12Mo and 130Cr12V1 were investigated. Various methods and technologies of electrode-arc surfacing, heat treatment, thermo-chemical treatment, plasma and electron-beam technologies for new surfacing steels, creating metastable states and bringing about the effects of self-organizing and self-strengthening at testing and exploitation are used in this work.

A new powdered welding wire with the shell of the low-carbon steel was designed, which contains a powder mixture of the following alloying materials: nitride low-carbon ferrochromium, ferromanganese, ferrotitanium, ferrovandium, silica-calcium in given proportions, which are inexpensive and available in Ukraine and many other countries. Variation of the chemical composition under laminated welding provides in the surfaced layer the reception of the metastable austenite-martensite alloy on Fe-Cr-Mn base with different correlation of the phases. The average chemical composition of the surfaced metal and investigated tool steels are given in Table 1.

Table 1

The average chemical composition of the surfaced metal and researched tool steels

steel grade	element concentration, mass %							
	C	Cr	Mn	Si	Ti	V	Mo	N
20Cr8Mn6SiNVTi	0.18	7.98	5.59	1.04	0.052	0.08	-	0.004
20Cr12Mn9SiNVTi	0.20	10.65	8.44	1.02	0.062	0.12	-	0.007
130Cr12V1	1.32	12.2	0.30	0.35	-	0.82	-	-
150Cr12Mo	1.48	12.3	0.34	0.32	-	0.20	0.55	-

The specimens were face-deposited with the designed powder wire of the grade PP-Np-12Cr12Mn9SiNVTi diameter being 4 mm on welding automatic machine of the models A1401 with power source “ВДУ 1001”. The electrode-arc surfacing was made on the carbon steel St. 3 (0.2 % C) by open arc at welding current 300...360 A, voltage of the arc 28...30 V. The first lot of the specimens was surfaced in one, in two, or three layers, according to the scheme, brought on Fig.1 and Fig.2.

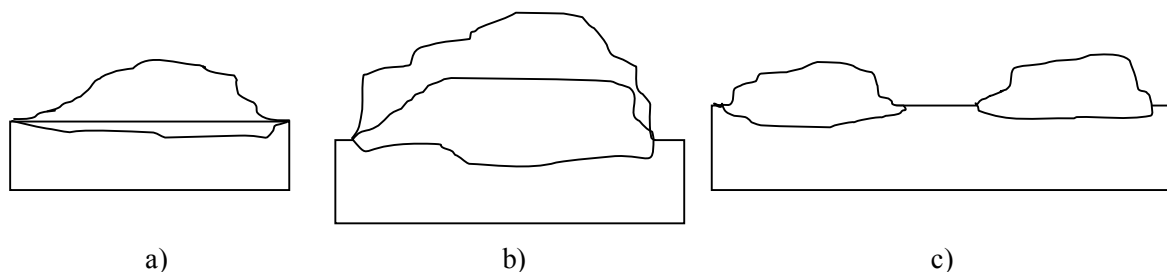


Fig. 1 – The schemes of surfacing: a) single-layered (№1); b) two-layered (№2); c) single-layered welding with the interval between beading fillets (№3.1 and №3.2, left to right accordingly)

From the surfaced metal a specimen for investigation was sliced and mechanical tests by size 10x10x55 mm. For determination of the phase composition of the specimens on x-ray diffractometer “ДРОН – 3” diffractograms were recorded in iron K_{α} – radiation in the range of angles 2θ 54...58 grad.

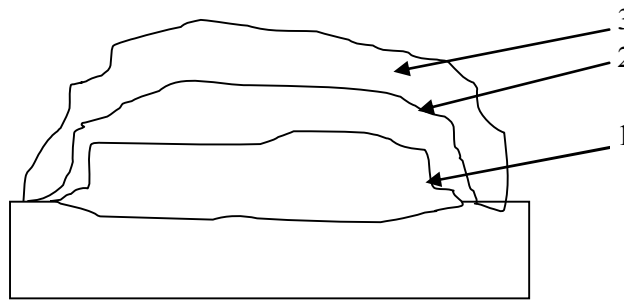


Fig. 2 – The scheme of the three-layer surfacing with interleaving of the layers of the surfaced metal: 1 - first; 2 - second; 3 - third

Plasma treatment of standard tool steels 150Cr12Mo and 130Cr12V1 (chemical composition shown in table 1) was carried out by means of plasmatron of indirect action, its design is described in [9]. Argon was used as plasma forming gas, its consumption being 1.0...2.3 m³/h at the current level equal to 360...400 A. The amount of heat was regulated by gas consumption and plasmatron motion velocity in relation to the specimens surfaces. Electron-beam treatment (EBT) for specimens of ample tool steels 150Cr12Mo and 130Cr12V1 was carried out on YL-185 electron-beam experimental and industrial plant, the amount of heat being regulated by the capacity of electron beam and velocity of its motion alongside the sample's surface.

The measurements of hardness of the specimens in surfacing condition and after heat treatment were taken on the Rockwell instrument with the load of 1500 N on cross metallographic specimens. The micro-hardness of the structural components (austenite, martensite, carbides) was measured on «ПМТ-3» instrument by indenter of the diamond tetrahedral pyramid with angle 136° under loading 1.96 N.

Tests on impact strength (impact energy) conducted on specimens 10x10x55 mm in size with U-notched specimen at the impact testing machine “ИО 5003” with maximum energy of the blow of the pendulum 300 J.

Optical metallographic method was made on microscopy “Neophot-21”. Tests of wear-resistance were made by two methods: at the dry friction sliding of metal at metal (ε_f) and at impact-abrasive wearing ($\varepsilon_{i.a.}$). Tests at the dry sliding friction were made at machine “МИ-1М” on scheme test sample - a roller (the checking body), revolving with constant speed 500 min⁻¹ (peripheral-linear speed in the friction zone is 1,31 meter per second). The time of wear-resistance averaged: quotient – weighting-to- weighting – 3 min, the same test time of chafing averaged 24 min. The specimens for wearing were of 10x10x27 mm size. Hardness of the steel roller was 25 HRC. Weighting was made up to ± 0.0001 gram. The load upon the specimen was 67 N.

Tests of the specimens 10x10x26 mm in size at impact-abrasive wearing were made on special installation [6] in the environment of molten cast-iron shot (the fractions 0,5...1,5 mm) at the speed of rotation of the samples 2800 min⁻¹. The same test time of average wearing 25 min. including the interval 5 min. weight-to-weight. 3 specimens undergo testing simultaneously.

The relative wear-resistance under these two types of wearing (ε_f , $\varepsilon_{i.a.}$) was defined on the formula:

$$\varepsilon = \frac{\Delta m_{st}}{\Delta m_s};$$

Δm_{st} , Δm_s - a loss of the mass of the standard and researches samples for equal test time.

As standard steel 45 (0.45 % C) was used in annealed condition by hardness 160...180 HB (for tests of samples of surfaced Fe-Cr-Mn steels), or tool steel 150Cr12Mo, after quenching by hardness 60...63 HRC (for tests of tool steels 150Cr12Mo and 130Cr12V1 after plasma or electron-beam treatment).

Results of experiments and discussion.

1. *Electrode-arc Surfacing Fe-Cr-Mn metastable steels.* Developed was the composition of new economical powder-like (nickel free) wire and investigated were the welding and surfacing characteristics for electrode-arc surfacing, the structure and mechanical properties of deposited metastable metal, capable of self-strengthening during the process of wearing.

For the overlaying surfacing by the powder wire it was provided well ganging of the built-up layer, without visual defects, pores and flaws. The average height of the built-up layer including the base at single-layered welding was estimated to be – 6...9 mm, for two-layer surfacing – 10...14 mm, and for three-layer surfacing – 15...18 mm. The average chemical composition of steel he steel (12...20)Cr(8...12)Mn(6...11)SiVTi in the surfaced metal was kept. Metal microstructure in the surfacing condition without heat treatment consists of martensite and austenite. According to the results of the X-ray structural phase analysis there are 15...25 % of the retained austenite (A_{ret}) in the single-layered surfaced metal. At the figure 3 are brought panoramic images of the structures of the overlaying surfacing of the first lot.

The structure is incomplete at the entire deposited metal. Martensite has package (rack) structure. The top zone extent 0.6...0.8 mm of the overlaying surfacing №1 (fig. 3) has fine-crystalline structure.

Hereinafter follows the zone with strongly pronounced stretching (mainly vertically) crystals, which grew up in the direction opposite to the heat-conducting path into the depth of the basic metal. This zone has the length of 2...5 mm. Further a conversion zone is situated adjoining the zone of alloying. Its thickness is 0,14...0,3 mm. Under it the thermal effect zone is found and then follows ferrite-pearlite structure of the basic metal.

The upper layer of the surfacing №3.1 (fig. 3 b) also has fine-crystalline structure, then follows a zone of the stretching crystals right up to the fusion and transitive zone, then follows ferrite-pearlite structure of the basic metal. It's up to the surfacing №3.2 that the fine-crystalline upper layer follows next layer with columnar crystals (the thickness beside 20 μm and length around 0.3...0.5 mm). Then we can see the zone of the fusion, the zone of the thermal influence and base metal.

The measurements of the hardness and microhardness were conducted in vertical (top-down) and horizontal directions. The hardness in the one-layered surfaced metal was 47...50 HRC, and decreased in the conversion zone till 37 HRC that matches to the martensite-austenite structure. In the fusion zone the hardness decreased till 15...19 HRC of the basic metal (St.3 (contain ~0.2 % C)). The change of micro-hardness in the one-layered surfaced metal is showed on the figure 4. Microhardness of the alloy with the martensite-austenite structure is 525...600 HV, and at the surface of the overlaying surfacing the hardness is a little less (520...550 HV) than at the fusion zone (580...620 HV). It is because of the larger content of the A_{ret} , as a result of the larger concentration of the alloying elements, that decrease martensite point M_s . Accord-

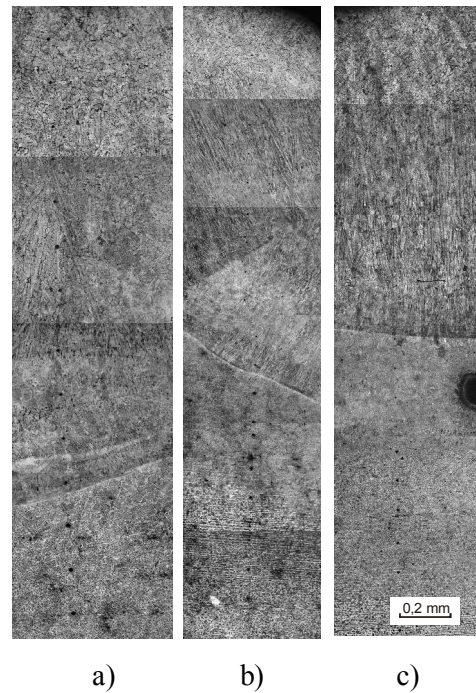


Fig. 3 – The microstructure of the surfaced layer of the single-layered overlaying surfacing (panoramic filming): a – overlaying surfacing №1; b – overlaying surfacing №3.1; c – overlaying surfacing №3.2

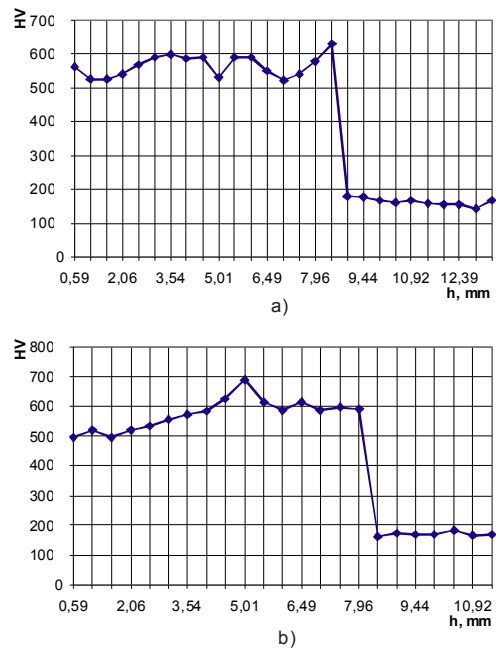


Fig. 4 – Distribution of the micro-hardness in vertical direction in the surfaced metal 20Cr8Mn6SiVTi: a – the overlaying surfacing №1; b – the overlaying surfacing №3.1

ingly at the fusion zone the surfaced metal is more mixed with the plain steel of the basis. The sharp reduction of the micro-hardness occurs in the conversion zone from 580...620 HV till 150...190 HV of the basic metal (St.3).

At the two-layer surfaced metal the change of the micro-hardness differs. The hardness of the upper layer is less than in the one-layer overlaying surfacing. This is possible because of the reception of the more alloyed upper layer that during the surfacing was mixed with low alloyed surfaced layer. Accordingly the microstructure two-layer surfaced metal was martensite-austenite with large content of A_{ret} .

It was important to study the influence of quenching on the structure and features of the surfaced metal. The samples indented of the three-layer surfacing were exposed temperature quenching 950 and 1150°C and were tempering at the temperature of 220°C, the stand-up at about one and half an hour (to decrease hardening stress).

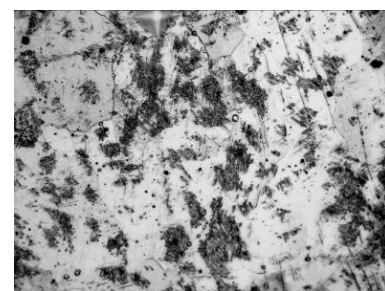
The microstructure of the upper layer is mostly austenite with content of the quenching martensite at about 35...45 %. The microstructure of the middle and lower layers is mostly martensite with content of martensite at about 70...85 % (rest A_{ret}).

After the quenching at the heightened temperature 1150°C the microstructure of the surfaced steel (12...20)Cr(8...12)Mn(6...11)SiVTi changes essentially (fig. 5). Upper and middle layers become practically completely austenitic with small presence ϵ -phases or defects of the packing. This is conditioned by dissolution carbide chromium ($Cr_{23}C_6$) and vanadium (VC) in austenite, by increasing of the contents of alloying elements in austenite that causes the reduction of martensite point M_s .

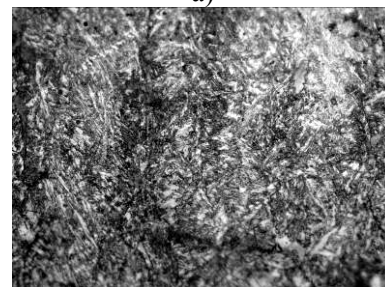
As the result the degree of stability of austenite increases. In the lower layer side by side with austenite appears quenching martensite. The gained microstructure corresponds to hardness of the surfaced metal.

The mechanical characteristics of three-layer metal in surfaced and quenched conditions were provided in table 2. From table 2 it follows that alloy (12...20)Cr(8...12)Mn(6...11)SiVTi in surfaced condition (without heat treatment) has the highest hardness (40 HRC), quite high relative wear-resistance in condition of dry friction, but the smallest impact strength and small impact-abrasive resistance ($\epsilon_{i.a.}$). This is because of the reception of mainly martensite structure (martensite transformation goes during the process of cooling with overlaying surfacing temperature), also because of the raised level of the stress, as well as small quantity of A_{ret} . The quenching at the temperature of 950°C reduces hardness till 35 HRC, simultaneously increases comparative wear-resistance in condition of dry friction and impact-abrasive wearing. This contradiction is explained by the reduction of the contents of quenching martensite in the structure of the surfaced metal and by increasing the contents of A_{ret} . At friction and impact-abrasive wearing influence in surface layer of the samples A_{ret} transformed into deformation martensite and causes self-strengthening of the surfaced metal. As A_{ret} is a more toughness phase and transformed in martensite in zone of the flaw development [8], increases also impact strength.

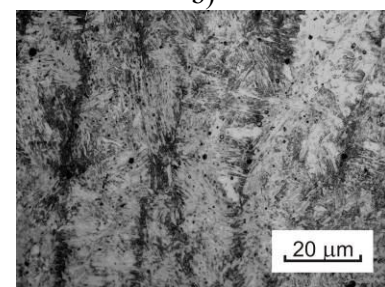
The metastable character of the surfaced metal and its ability to self-strengthening in surface layer in process of wearing under the action of wearing environment, due to realization of $\gamma \rightarrow \alpha'$ DIMIT – is the most important particularity and advantage over similar deformation stable materials. The self-organizing and self-strengthening effects are conditioned by formation of more strength and hard martensite deformation that is accompanied the process of micro-stress relaxation



a)



b)



c)

Fig. 5 – The microstructure of the three-layered surfaced metal (12...20)Cr(8...12)Mn(6...11)SiVTi after the quenching at the temperature of 950 °C, tempering 220 °C: a – upper layer; b – middle layer; c – lower layer

[3]. In addition, DIMIT becomes the powerful mechanism of the absorption and redistribution of mechanical energy of the external influence on material, in this connection it's considerably less part remains on destruction of the material [8].

Table 2

The Mechanical characteristics of three-layer surfaced metal 12...20)Cr(8...12)Mn(6...11)SiVTi

Condition of the surfaced metal	HRC	Relative wear resistance (dry friction), ε_f	Relative impact-abrasive wear resistance $\varepsilon_{i.a.}$	KCU, J/sm ²
Without heat treatment	40	3.8	4	7.0
Quenching at the 950°C, tempering 220°C	35	4.2	6.5	8.5
Quenching at the 1150°C, tempering 220°C	32	1.5	9.8	15.4

The hardness of the surfaced metal decreases more when quenching temperature rises from 950 to 1150°C, impact energy increases twice (KCU=15.4 J/sm²), vastly increases impact-abrasive wear-resistance ($\varepsilon_{i.a.}$). This is because of the becoming of the $\gamma \rightarrow \alpha'$ structure mostly austenite, consequently increases the possibility and volume of $\gamma \rightarrow \alpha'$ DIMIT during the wearing tests and also self-strengthening effect in the surface layer. However, relative wear-resistance in condition of friction metal on metal in this case falls because DIMIT develops in vastly smaller degree, and does not provide necessary self-strengthening of the surfaced metal. The reason of increasing the content of austenite in the structure of surfaced steel (12...20)Cr(8...12)Mn(6...11)SiVTi is dissolution of chrome carbides in austenite with increasing of the temperature of the quenching, and increasing of alloying degree in austenite that lowers the critical point M_s .

So, for different conditions of the usages of surfaced parts it is necessary to select the mode of quenching, providing optimum phase composition (the correlation of martensite and austenite), the degree of austenite metastability: for conditions of dry friction of the metal on metal temperature of quenching must be 950°C, and for impact-abrasive wearing - 1150°C.

Hence, during operation of the metastable alloy surfaced by the developed powder wire effectively used are internal reserves of the material itself to raise mechanical and official characteristics due to the realization of the self-organizing and self-strengthening effects at friction and wearing in consequence of $\gamma \rightarrow \alpha'$ DIMIT.

2. *Plasma and electron-beam treatment tool steels.* Application of plasma and EBT has been widely used lately for strengthening of a wide range of parts and tools [8-12]. Treatment with application of the sources of concentrated energy (SCA) makes it possible to achieve increased amounts of residual austenite (A_{res}) inside the deposited layer of the bulk of steel and cast-iron grades under treatment [8-12], thus opening some possibilities of forming and using meta-stable states in order to improve their properties.

Plasma treatment, depending on the mode and the amount oh heat applied caused local heat of the surface without its melting, or its remelt with different degree of melting. EBT caused melting of specimens surfaces with greater depth (4...6.5 mm), as compared to plasma treatment.

In both variants high-speed surface heating was performed ($V_{heat}=10^3 \dots 10^5$ °C/sec) after that there followed quick natural cooling, without application of cooling media, due to heat application inside samples' depth.

Microstructure of tool steel 130Cr12V1 is summarized in Fig. 6. In its initial state it is characterized by carbide heterogeneity (Fig 6a), with chromium carbide distribution alongside grain boundaries, which usually cannot be eliminated by heat treatment and facilitates brittle destruction of tools edges. Plasma treatment and EBT with surface melting eliminate the original carbide non-uniformity at the depth equal to 0.4...2.5 mm. Some special equiaxial cellular structure of oversaturated austenite with slightly bigger hardness (500 HV) than equiaxial austenite (Fig. 6b) is observed directly near the surface of the melted area. Then alongside the depth there is an area of a specific columnar crystals, their axes are situated normally with regard to the melted surface, this showing heat removal from the surface towards in depthward direction of the samples at accelerated cooling, corresponding to the ve-

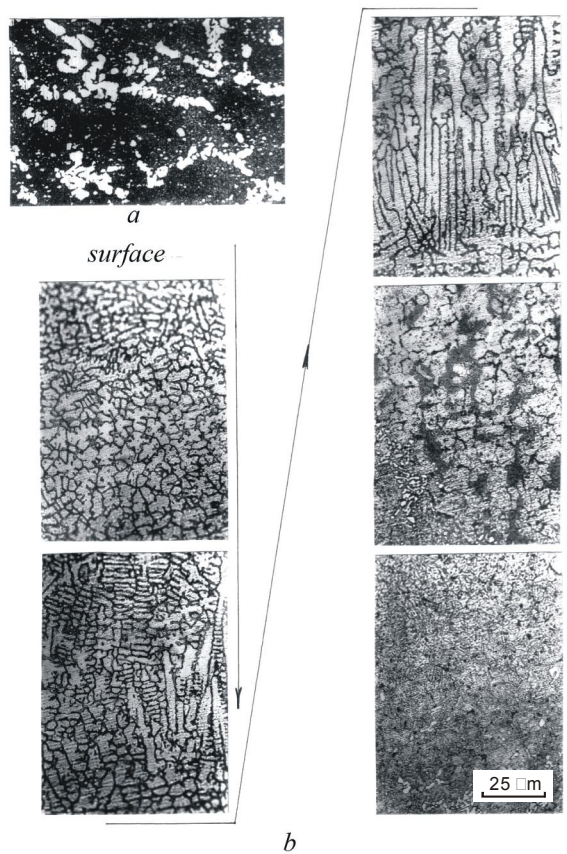


Fig. 6 – Microstructure of steel 130Cr12V1 initial (a) and after electron-beam treatment with melting (b)

locity of their growth. Inside the area of plasma or electron-beam melting the crystallites sizes are substantially smaller than grain sizes of the original structure: inside the area of plasma melting or EBT the crystals had widths equal to $\sim 3 \mu\text{m}$ and length $7 \dots 20 \mu\text{m}$, while the average diameter of martensite grains prior to treatment was $\sim 40 \mu\text{m}$. Fine carbide capsule was found alongside the boundaries of columnar crystals (Fig. 6b) Austenite grains acquire regular shapes as their distance from the surface increases, while the thickness of carbide capsule becomes slightly bigger. The increase of intensity and duration of plasma action makes the thickness of melted area bigger by 0.4 to 2.5 mm, reducing its hardness from 61 HRC to 38 HRC, thus proving the formation of mostly austenite (oversaturated) structure of the surface layer.

Gradually, austenite-carbide structure of the melted area is transformed into the area of thermal influence (ATI), consisting of austenite, martensite and carbides (Fig. 6b). Alongside ATI depth the amount of martensite gradually increases, while the amount of A_{res} is reduced. After ATI the structure is gradually transformed into the original ferrite-carbide mixture.

Relative wear resistance of 150Cr12Mo steel grade, under conditions of dry friction (ϵ_f) is summarized in Table 3 and it depends upon the structure obtained after plasma treatment.

Table 3

Influence of plasma treatment upon hardness and wear resistance of 150Cr12Mo steel grade

№ of the mode	Degree of plasma action	Melt depth, mm	HRC	Value of ϵ_f	Impact Energy KC, J/sm ²
1	Heating without melt	-	60	0.94	6.5
2	Micro-melting	0.8...1.7	58	1.0	7.8
3	Average degree of melting	2.5...3.0	61	1.32	12.0
4	Strong degree of macro-melting	3.2...4.4	38	0.8	9.5
-	Control quenching (with furnace heating 1030°C), tempering 200°C	-	63	1.0	6.0

Plasma treatment with medium macro-melting improves relative wear resistance under the conditions of sliding friction (ϵ_f) by 1.32 times, as compare to standard quenching for martensite-carbide structure of maximum hardness 63 HRC. The reasons of such increase in wear resistance may be formation of specially oriented cellular carbide framework and transformation of oversaturated austenite into martensite during the process of wear at optimal kinetics. Besides, the bulk of mechanical energy supplied to a specimen is consumed for $\gamma \rightarrow \alpha'$ DMT and accordingly, its lesser part remains for destruction of mechanical bonds and metal separation [8].

Plasma treatment without melting and with micro-melting, when only small amount of A_{res} is formed and strong macro-melting with a big depth of the area of oversaturated, over-stable austenite does not give any advantage for chosen conditions of wear. Elimination of carbide non-uniformity (see Fig. 6a) and formation of fine grain structure of austenite, inserted into the fine carbide framework

increases impact energy by ~ 1.4 times from $6 \dots 7 \text{ J/cm}^2$ to 12 J/cm^2 , reducing the probability of a brittle chip. This may be considered as an indirect factor of increase of service lives and wear resistance of parts and tools.

Conclusions

1. The created powder surfacing wire provides good fusing of surfaced metal with steel base between layers at two- and three-layer overlaying surfacing. Received economical alloyed metal of the composition $(12 \dots 20)\text{Cr}(8 \dots 12)\text{Mn}(6 \dots 11)\text{SiVTi}$ in surfaced condition has martensite-austenite and austenite-martensite structure with metastable γ -phase able to the deformation induced $\gamma \rightarrow \alpha'$ martensite transformation conversion during wearing.
2. In the microstructure on overlaying surfacing section the relationship between austenite and martensite changes that influences upon metastable degree of γ -phase.
3. The temperature of quenching lets to adjust the quantitative correlation between austenite and martensite in the surfaced metal that defines shaping of mechanical characteristics due to the change of metastable degree of austenite.
4. Plasma and electron-beam treatment, depending upon the amount of heat application cause formation of columnar (at melting) or fine disperse (when being heated without melting) microstructure with increased content of metastable residual austenite in high-chromium tool steels.
5. After rational modes of plasma and electron-beam treatment an optimal martensite – austenite-carbide structure is formed, ensuring inside the strengthened surface layer $\gamma_{\text{res}} \rightarrow \alpha'$ DIMT during subsequent wear. This increases substantially relative wear resistance of 150Cr12Mo and 130Cr12V1 steels, eliminating carbide non-uniformity of cast metal structure and increasing impact energy by 1.5...2 times, reducing the probability of metal's brittle chip.

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Received 06.11.2013