MODIFICATION OF HARD ALLOY WC--STEEL 110G13 BY A PULSED LOW-ENERGY, HIGH-CURRENT ELECTRON BEAM

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Using metallography, x-ray diffraction analysis, diffraction electron microscopy, and microhardness measurements we have investigated the results of the action of a pulsed low-energy, high-current electron beam on the phase composition, defect structure, and mechanical properties of the hard alloy WC-30% steel 110G13. We have observed and studied in detail the regions of elevated microhardness (the microhardness of the material in these regions is 1.5-2.0 times greater than the original value) located on the irradiated surface and in the interior of the material. We have shown that the number of zones of elevated microhardness increases with an increase in the number of pulses in the electron beam treatment. We consider the mechanisms for hardening of the material by a low-energy, high-current electron beam. We conclude that the increase in the microhardness of the alloy is due to dispersion of the carbide phase, precipitation of nanometric complex carbide particles, strain hardening, and hardening due to polymorphic transformation ($\gamma \rightarrow \varepsilon$) of the binder.

In a series of papers [1-5] describing work done on structural steels, we have shown that pulsed electron beams are able to exert a controllable modifying action on the structure, phase composition, and consequently the mechanical characteristics of materials.

The goal of this work is a structural and phase investigation of the hard alloy WC--steel 110G13 subjected to treatment by a pulsed low-energy, high-current electron beam. In contrast to structural steels, this alloy is a clear representative of structurally inhomogeneous materials and consists of tungsten carbide crystallites of size 2.2 $\mu$m, separated by the plastic binder, steel 110G13 (30% of the weight of the alloy). The binder phase was found in the metastable austenitic state, formed by quenching at 1520 K into potassium nitrate. The specimens were taken in the shape of disks of 25 mm diameter and 8 mm thickness. Irradiation was done under a vacuum of $10^{-5}$ mm ($\approx 0.0013$ Pa) on a high-current electron accelerator under the following conditions: maximum electron energy, $E = 40$ keV; pulse duration, $\tau = 2.5$ $\mu$sec; energy density, $W = 40$ J/cm$^2$. We considered two series of specimens with the number of pulses equal to $N = 1$ and 10, respectively. In this case, we used metallography, x-ray diffraction analysis, and diffraction electron microscopy of thin foils. The mechanical properties were investigated by measuring the microhardness of both a frontal section (relative to the low-energy, high-current electron beam) of the specimen and a section perpendicular to it according to GOST 8.398-80.

As shown by the previous investigations [6], in the original state the hard alloy WC--steel 110G13 is mainly a mixture of two phases: tungsten carbide (simple hexagonal crystal lattice, $a = 0.291$ nm, $c = 0.284$ nm) [7] in the form of crystallites of micron size; and steel 110G13 (0.3 of the weight of the material), found in the austenitic state [8] and playing the role of a plastic interlayer. Along with these phases, secondary phases were present in insignificant amounts: tungsten carbides $W_2C$ (hcp crystal lattice) and $WC_6$ (cubic lattice), and also complex carbides of the type $M_{12}C$ and $M_{23}C_6$ based on iron and tungsten.

The results of the metallographic investigations of longitudinal and transverse sections of the hard alloy treated with a low-energy high-current electron beam are presented in Fig. 1. We clearly see that the action of the electron beam first of all leads to melting of the surface layer of the material (Fig. 1a) and, secondly, to the appearance of macrocracks due to very high surface heating rates ($\approx 10^{11}$ K/sec) and cooling rates ($\approx 10^9$ K/sec) [2]. The depth of the melted layer and the density of the cracks appreciably increases (by a factor of two), with an increase in the number of pulses for the electron beam treatment from 1 to 10.

Fig. 1. Macrostructure of the hard alloy WC–steel 110G13, subjected to multiple-pulse \((N = 10)\) treatment with a low-energy, high-current electron beam.

Fig. 2. Microhardness \(HV\) of the hard alloy WC–steel 110G13 vs. distance to the surface treated with a pulsed low-energy, high-current electron beam: curve 1 — \(N = 10\) pulses; curve 2 — \(N = 1\) pulse.

The microhardness of the alloy, as a function of the number of pulses in the electron beam treatment and the distance from the surface of the specimen, is represented in Fig. 2. From the presented results it follows that treatment with a low-energy, high-current electron beam promotes firstly, formation of a hardened surface layer and, secondly, creation of zones of elevated microhardness which are located quasiperiodically over the depth of the material. Increasing the number of pulses from 1 to 10 leads to an increase in the number of microhardness peaks and their quasiperiodic location over the depth of the material (with period \(\approx 500 \mu m\)). We note that these effects have been observed previously in [1-5].

Based on the results obtained by optical microscopy and the microhardness tests, further efforts were directed toward investigations (using x-ray diffraction analysis and diffraction electron microscopy) of the structural and phase state of the surface layers and the layers lying in the region of the first microhardness maximum.

The x-ray diffraction investigations, done on a DRON-UM1 instrument with filtered copper radiation, revealed a substantial change in the phase composition of the irradiated near-surface layer. Specifically, the major structural component (independently of the number of treatment pulses) is the high-temperature modification of tungsten carbide \(\beta-W_2C\) (hcp crystal lattice, \(a = 0.292\ nm, c = 0.433\ nm\)). The amount of the original carbide WC does not exceed 5%. The binder phase (steel 110G13) is found in one-phase \((N = 1\) pulse) or two-phase \((N = 10\) pulses) states (Fig. 3, curves 2 and 5 respectively). Etching the surface of the material irradiated once and subsequent x-ray diffraction analysis showed that at a depth of \(\approx 10\ \mu m\) (the dip on the microhardness curve in Fig. 2), the phase state practically corresponds to the original state (Fig. 3, curve 3). At a depth of 25 \(\mu m\) (the microhardness peak in Fig. 2), we see a change in the phase state of both the binder (Fig. 3, curve 4) and the carbide phase (we observe high-temperature modifications of tungsten carbide).

The electron microscopy investigations of the near-surface layer of the specimen (a layer of thickness 0.15-0.20 \(\mu m\), including the surface itself) after multiple \((N = 10\) pulses) electron-beam treatment revealed the following — the surface layer of thickness 0.10-0.15 \(\mu m\) is a mixture of two types of substructures: nanometer-size crystallites (\(d \approx 40\ \mu m\)) and submicron-size crystallites (\(d \approx 0.2\ \mu m\)). In this case, we should assume that the first of these substructures forms a layer making up the surface, the second lies somewhat deeper. This hypothesis is based on the fact that we observed a nanometric structure in foils obtained by one-sided thinning (thinning on the side opposite to the surface treated with the low-energy, high-current electron beam). The submicron structure was observed in foils which had undergone a 5-minute surface polishing with an ion beam, which corresponds to removal to a depth in the material of \(\approx 0.1\ \mu m\).

The nanometric structure, as follows from Fig. 4a, is formed by teardrop-shaped single crystals separated by interlayers whose average thickness was \(\approx 8-10\ \text{nm}\). Diffraction analysis shows that the crystallites are a high-temperature cubic modification of tungsten carbide \((\text{WC}_{\alpha})\); the interlayers are metallic tungsten of the high-temperature modification \((\beta-W)\) (Fig.