INFLUENCE OF HEAT AND LASER TREATMENT ON THE STRUCTURE AND PROPERTIES OF R6M5 HIGH-SPEED STEEL

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ABSTRACT

The influence of laser treatment on the structure, microhardness and residual stresses in the surface layers of high-speed tool steel R6M5 is investigated. It is shown that laser treatment, carried out as a finishing operation of thermal hardening, can significantly change the structure and properties of steel R6M5. The modes of hardening treatment, which provide an improved combination of properties, due to the uniform distribution of residual austenite and martensite, are determined. This helps to increase the endurance of steel R6M5 when exposed to its surface pulsating contact stresses.

Keywords: high-speed steel, laser modification, structure, microhardness, residual stresses, contact wear.

INTRODUCTION

Laser processing of high-speed steels is one of the methods to improve the tool performance but requires precise control of parameters to obtain the desired properties and structure of the surface layer [1, 2]. This is due to the increased solid solution alloying which helps to stabilize the austenitic phase. The growth in the proportion of metastable γ-phase, on the one hand, allows increasing the viscosity of the material, on the other one, reduces the hardness and increases the tendency to dynamic aging that often leads to rapid nucleation of internal defects deteriorating the properties of the surface layer [3]. These features make urgent issue the investigations of operating characteristics of the modified layers of high-speed steels after laser treatment, as well as other operations that are able to adopt the final structure to external impact.

EXPERIMENTAL

The surface layers of high-speed steel R6M5 are the objects of research. They have been modified by laser irradiation after the traditional heat treatment aiming the formation of martensite structure with hardness equal to 61-63 HRC [4, 5]. Fiber ytterbium laser with a wavelength of 1070 nm and radiation power 1.3 kW was applied. The scanning laser beam with a frequency of 220 Hz in the transverse direction provided the formation of quasi-stationary heating zone with dimensions 0.7×6 mm. Longitudinal movement along the treated samples' surface was carried out by machine at speeds from 600 to 1350 mm/min. The most low movement speed - 600 mm/min was chosen empirically to achieve the effect of partial surface melting. Processing each of further set was carried with increasing speed for 150 mm/min. Modification was performed both as the final processing and as with subsequent heat treatment (high temperature tempering at 560ºC) aiming the dispersed hardening of austenite. Metallographic analysis of diffusion layers at all stages of the study was conducted with optical microscope METAM PB22. Microhardness was measured using the device PMT-3 at a load of 2N. A test of contact wear was performed applying the original unit [6] which provides a contact loading of end surface of the flat part of the sample due to its rolling without slipping along a working surface of the disk counter-body. The study of the residual stresses distribution was carried out according to the method proposed by M.M. Saverin [7].

RESULTS AND DISCUSSION

After heat treatment the microstructure of steel R6M5 consists of the martensite, residual austenite and carbide inclusions. The laser hardening resulted in formation of a bright, almost non-etching region on the surface of all samples. The minimum thickness of the layer was up to 20 μm after treatment with the greatest velocity equal to 1350 mm/min. At laser beam velocity of 600 mm/min this thickness reached 150 μm. The distribution of microhardness through the cross section of the samples with the thickest layer was characterized by non-monotonic dependence (Fig. 1(a)) and it increased from 10 GPa on the surface to 13 GPa in the sublayer at depth of about 700 - 800 μm. Then microhardness gradually decreased to values of the core. Similar dependences were observed in the samples processed at speeds of 750 mm/min and 900 mm/min. The surface layer contained signs of overheating and had poor etching ability in an alcohol solution of nitric acid indicating the predominance of austenite in its solid solution (Fig. 2(a)). Microhardness' distribution curve changed its character at depths of 500 μm - 600 μm for samples treated with higher velocities. The abrupt lowering of microhardness was checked indicating the processes of self-tempering.

The use of high-temperature tempering as an operation after laser treatment led to the removal of the marked differences in the distribution of microhardness giving all dependences the traditional view of hardened layers (Fig. 1(b)). In the structure of the hardened layer changes associated with the precipitation-hardening of residual austenite took place but part of austenite remained unchanged (Fig. 2 (b)).

The value of residual stresses formed in the hardened layer does not exceed 200 MPa. From Fig. 2 it is seen that in the surface layer subjected to the most severe thermal influence the value of the stresses is not more than 70 MPa. In samples with a bigger thickness of the austenite layer and with a strong overheating structure compressive stresses occur and in samples with thinner layers stretching stress forms. Maximum stresses values are observed at a depth of about 1.5 - 2.0 mm and they have compressive nature. These stresses have higher value for thicker laser layer. Their appearance is due to a compensatory effect in response to tensile stresses that are spread to a depth of 0.5 mm for all layers. Low stress values are probably due to the relaxation ability of the residual austenite. Conducting high tempering after laser treatment resulted in formation of compressive stresses in a sublayer, which together with the analysis of the microstructure of the hardened layer confirms the realization of partial phase transformation in the austenite (Fig. 3).

The results of samples testing under the action of pulsating contact stresses with an amplitude of 1300 MPa on their surface layer showed that at the initial stages microcracks formed in sub-surface layer of material at a depth of 0.1 - 0.2 mm (Fig. 4(a)). Since the surface layer has experienced the influence of the pulsating stresses the region of crack nucleation in the field of action of tangential stresses' maximum value appear at a depth of 0.2 mm or more [8, 9]. The local stress concentration is intensified by the boundaries of large grains. Due to overheating and oxidation they are the most fragile phase of the material. Crack propagation along the grain boundaries causes their deformation and the formation

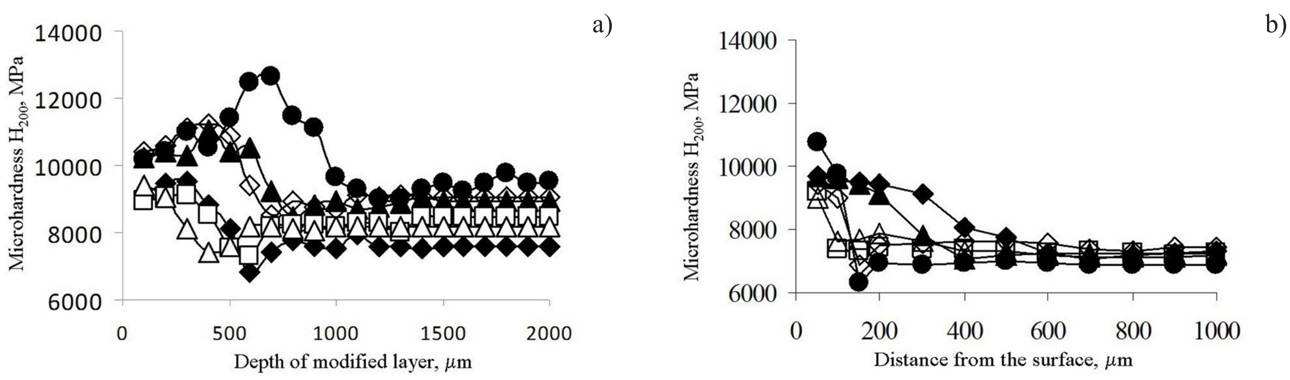
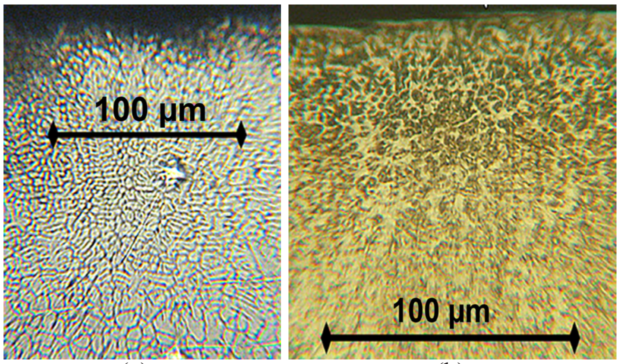


Fig. 1. Distribution of microhardness through the cross section of R6M5 steel samples for sets processed with the speed, mm/min: ● - 600, ◊ - 750, ▲ - 900, ♦ - 1050, □ - 1200, Δ – 1350: a) after laser modification; b) laser modification + tempering at 560ºC.



(a) (b)

Fig. 2. Structure of the surface layer of R6M5 steel samples subjected to laser processing at a speed of 600 mm/min as a final operation (a) and after additional tempering at the temperature 560ºC (b).

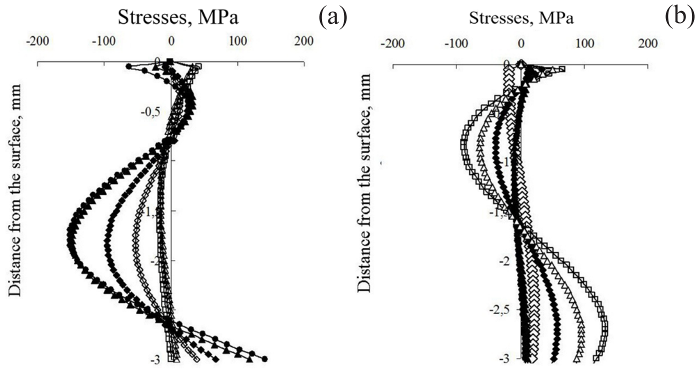


Fig. 3. Distribution of residual stresses in the modified by laser treatment of surface layer of R6M5 steel at velocities: ● - 600, ◊ - 750, ▲ - 900, ♦ - 1050, □ - 1200, Δ – 1350: a) laser modification; b) laser modification + tempering at 560ºC.

of sufficiently large internal cavities. Such mechanism of destruction of the subsurface metal layer does not depend on the presence in the alloy of the excess phases and can be realized in a homogeneous material [10, 11].

The wear of sample's contact surface run according to classical mechanism combining the growth of crack-like defect and its access to the surface [12]. The revealed morphological features reflect the mechanism of fatigue fracture in conditions of inter-crystallite development of corrosion [13]. That is apparent to the outer layer of alloy which has signs of overheating in the form of melted grain boundaries. After the running-in stage the destruction of the surface layer proceeds due to the formation of semi-circular cracks which cover the area of plastic deformation in the vicinity of the contact stripes. Changes of textural structure of the alloy in the vicinity of the contact fatigue crack are observed. Carbide inclusions occupy an equidistant arrangement with respect to the crack growth trajectory. Outside of the deformation area structural changes are expressed to a lesser degree. The moment of separation of the hardened metal fragment on the wear curves is reflected by stepped growth. Then the hardening of the deeper layers of metal begins. Wear of all laser modified samples run without significant differences despite the differences in the thickness of the modified layers. Usually one crack is involved in the pitting formation. The wear intensity of all samples sets having modified layer thickness from 0.7 mm to 1.2 mm is characterized by a similar dependency.

The samples after high temperature tempering are characterized with higher wear resistance. Wear curves have clear flat areas that reflect the period of the precision resistance of the surface layer (Fig. 5(b)). The wear rate in this period has a minimum value. Hardening of the surface layer is accompanied by changes in the structure. All components have oblong shape lying along the contact surface (Fig. 4(b)). Energy of external influence is dissipated due to the deformation of the structural components and the formation and propagation of contact fatigue cracks is hampered. Samples with a minimum thickness of the modified layer formed at a speed of 1350 mm/min showed best resistance.

(a)

(b)

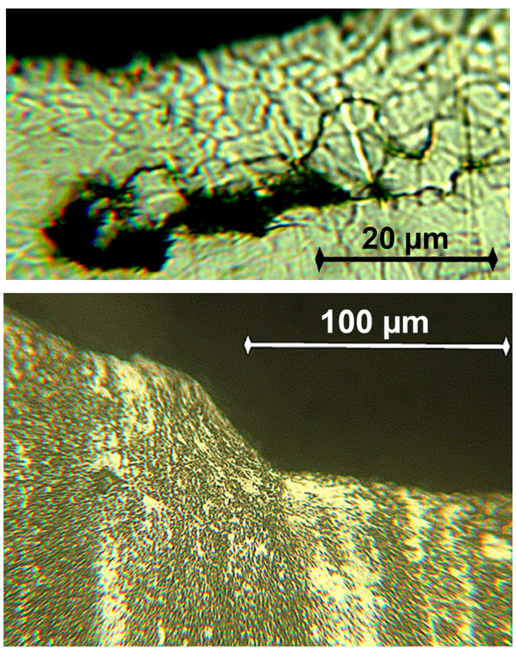


Fig. 4. Defects in the structure of the surface layer formed during the contact wear of the working surface of the samples modified by laser processing (a), and further subjected to high temperature tempering (b).

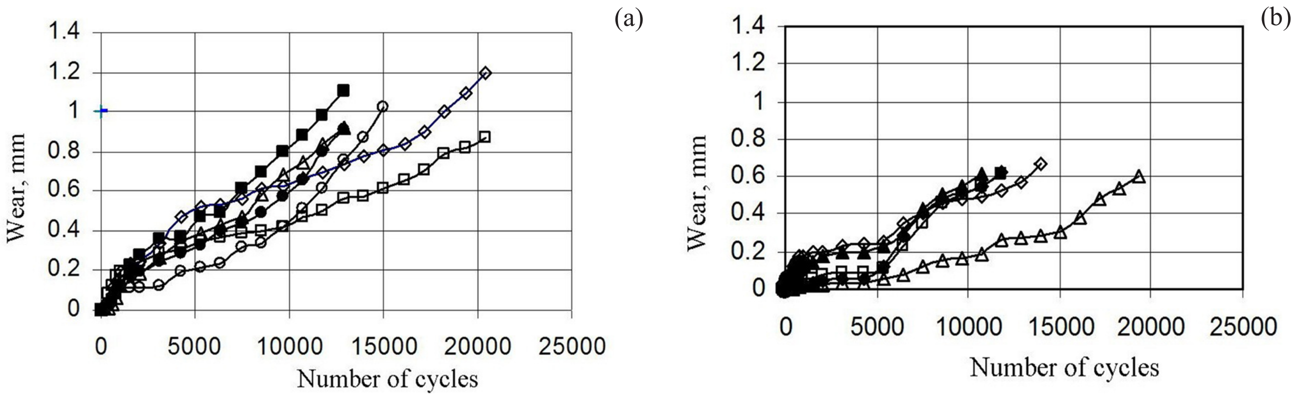


Fig. 5. Curves of the surface layer wear of R6M5 steel samples modified by laser processing (a) and after additional treatment with high temperature tempering (b).

CONCLUSIONS

Effects of laser hardening on the structure, microhardness and magnitude of residual stresses in surface layers of R6M5 high-speed tool steel are examined. It is established that the increase of laser scanning speed leads to a reduction in areas of structural transformations and the thickness of the layer with higher microhardness. It is noted that the exposure to high tempering at 560ºC leads to a change in residual stresses in the surface layer from tensile to compressive. Moreover, depth of microhardness distribution is reduced due to the elimination of the inner zone of self-tempering during the dispersion hardening of the material. As a result, the modified layer acquires an improved combination of properties due to uniform distribution in the solid solution of two phases: residual austenite and martensite. The best balance of mechanical properties occurs after laser modifying of the surface at speed of 1350 mm/min and a subsequent high-temperature tempering at 560°C. This contributes to increase of the period of the precision resistance of R6M5 steel under the influence on its surface pulsating contact stresses with amplitude of 1300 MPa.

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