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The microstructural aspects of abrasive wear resistance in composite electron beam clad coatings

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ABSTRACT

The effect of microstructure and phase composition of composite coatings based on manganese-containing or nickel-containing austenitic steel and containing either 10 wt.% TiC or 15 wt.% WC on the abrasive wear resistance has been studied. Both post-deposition heat treatment and self-aging in the course of deposition served to improve the relative wear resistance coefficient by 38–42 and 5–12% for Fe-20%Mn-4%V-4%Mo + 15%WC and Fe-20%Ni-4%V-4%Mo + 15%WC coatings, respectively.

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1. Introduction

High-energy beam processes are widely used now for depositing multi-functional both hard and soft coatings [1,2]. The hard coatings usually contain hard particles and it is not an easy task to obtain a sparingly alloyed composite coating with homogeneous distribution of these particles using such a process. Such a difficulty becomes even greater when there is a necessity of obtaining the nanostructured coatings [3,4]. Both phase content and particle size are changed during the cladding [5–7]. The driving force behind such a structural modification is the high ratio between the particle's surface area and its volume. At the same time, wear resistance of composite bulk materials is known to improve with refining the carbide grains [8,9]. Such an effect is achieved due to reducing the intercarbide binder layer's thickness, improving the mechanical interlocking of carbides and diminishing the risk of forcing out the binder during either abrasive or sliding wear. As shown [6–10], the nanosized grain WC–Co cemented carbides made by sintering really show the improved wear resistance as compared to that of their standard versions.

However, when it comes to high-energy beam depositing a coating to be reinforced by hard carbide particles, the coating's structure

may acquire defects such as (i) inhomogeneous distribution of carbides across the coating's thickness and/or (ii) the carbide network. The first type of the defect structure arises due to a big difference in specific volumes between the carbides and binder while the second type is formed under condition that the primary carbide is capable of dissolving in the melted matrix and then precipitating in the form of secondary carbides. Both types of defect structure will serve to facilitate the predominant removal of the binder during the wear followed by dislodging of the carbides and loss of the wear resistance.

Attempts are known to deposit a WC–Co coating with a bimodal carbide size distribution using the HVOF method [11]. It has been found out that in spite of partial degradation of the phase composition during the deposition both fine-grained and nanostructured coatings demonstrated almost the same wear resistance which was twice as low as that of the bimodal coating. This example demonstrates us the importance of obtaining a coating which comprises the carbides of several size ranges homogeneously distributed in the bulk of the coating for improving its performance.

Next promising approach [12–15] provides for using a filler containing either TiC or WC carbide particles which have different solubility in the melted matrix. The points here are that (i) fine carbides will precipitate during solidification, (ii) oversaturated solid solution of carbide-forming elements in the matrix will be formed during fast cooling, (iii) even more fine carbides will precipitate from the solid solution in the bulk of the coating in

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the course of aging, (iiii) some amount of primary carbides will retain.

We have reported earlier [14] that multi-pass electron beam deposition process creates conditions for thermocycling the deposited metal what allows forming a multimodal carbide particle size distribution in the bulk of the coating irrespective of the austenite-forming metal used. Austenite is known to be an efficient binder for securing the carbide particles in the matrix due to its higher ductility as compared to other structural components. Also austenite is capable of partial strain-induced phase transformation hardening and thus it contributes to overall hardening of the composite [16,17]. Additional hardening may be achieved through alloying the matrix by such strong carbide-forming metals as vanadium and molybdenum. Vanadium is capable of forming fine carbides at elevated temperatures whereas molybdenum may be a part of mixed carbides [18].

The objective of this study is the effect of carbide particle size distribution on wear resistance of austenitic composite coatings deposited using the electron beam process.

2. Materials and research methods

2.1. Deposition

Powder formulations such as (Fe-20%Mn-4%V-4%Mo)+10%TiC (7.36 kg/cm³), and (Fe-20%Ni-4%V-4%Mo)+10%TiC (7.68 kg/cm³), (Fe-20%Mn-4%V-4%Mo)+15%WC (8.81 kg/cm³) and (Fe-20%Ni-4%V-4%Mo)+15%WC (9.13 kg/cm³) have been used as filler mixtures for electron-beam deposition in a chamber at residual gas pressure not higher 0.013 Pa. These formulations were prepared by blending together the powder components, compacting and sintering them. The resulting samples have been milled and finally sized into fractions. Powder fraction 50–350 μm was used for electron beam deposition on flat 30 × 30 × 200 mm substrates machined from low carbon 0.30 wt.% C steel. An electromagnetic powder feeder provided the feeding rate 40 g/min. No carrying gas was used.

Electron-beam deposition was carried out by feeding the powder mixture into a ~100 mm² melted pool generated by incident electron beam scanning linearly over the metal substrate surface [13]. To obtain the full coating area, the substrate was being moved inside the vacuum chamber with respect to a fixed electron gun and a powder feeder. More details on the deposition process are given elsewhere [12].

Since the substrate top layer was partially melted during the first pass, the bottom part of the coating was diluted by the deposited elements. The required thickness of coating was achieved by increasing the number of electron beam passes over the previously clad layers. On each pass, a portion of composite powder mixture has been frozen into the pool. The first pass provided melting to the depth of 0.6–0.8 mm whereas successive passes melted the pre-clad layers to the depth of 0.4–0.5 mm. Also the pre-clad layer or the substrate suffered partial melting in the process what has resulted in developing the heat affected zone (HAZ).

The density of all samples has been determined by cutting out a cube and measuring its mass and volume. Additionally, these samples have been examined for porosity using an optical metallography.

2.2. Structural analysis

Optical microscope Olympus GX 51 attached with SIAMS 700 analyzer and SEM instrument Philips SEM 515 attached with EDAX

ECON IV microanalysis peripheral equipment have been applied for structural characterization of the coatings and their worn surfaces.

Both transversal and longitudinal microsections have been prepared by mechanical grinding and polishing with diamond pastes of different grades. Chemical etching by 4% solution of HNO₃ in ethyl alcohol as well as in chloroazotic acid (1 part HNO₃ + 3 parts HCl) has been used to reveal the microstructures. Quantitative characteristics of the microstructure including number, size, shape and distribution of various phases were determined using the SIAMS software package.

XRD instrument DRON-7 operated in scanning mode in the 2θ range 15 to 150 degrees by 0.1 degree step and filtered CoKα-radiation has been used to identify the coatings' phases. Quantitative phase analysis has been performed by integrated intensities of the diffraction peaks.

2.3. Microhardness

Microhardness numbers (H_{μ}) of both the coating and the substrate at the depth 2 mm below the coating were measured using PMT-3 microhardness tester (ISO 4516:2002) operated at 100 μm step between the indents and load of 0.981 N. The measurements were carried out using two parallel indentation paths shifted by 50 μm in depth with respect to each other. The distance between the paths was 200 μm. Such a procedure allowed obtaining the microhardness in-depth profiles of 50 μm step in the coating.

2.4. Wear test

Method of loose abrasive particles has been used to determine the abrasive wear resistance of obtained coatings [12]. Medium carbon 0.45%wt. C steel samples annealed to 190–200 HV served as reference samples in wear tests. Angular quartz sand and electrocorundum particles of the mean size d_m in the range 160–350 μm were fed into the zone between the 50 mm diameter rubber roller and the sample which were pressed against each other by the force 44.1 ± 0.25 N. The rubber roller's sliding linear speed was 9.4 m/min. Dimensionless relative wear resistance ε was determined according to the expression as follows:

$$\varepsilon = \frac{\Delta m_r \rho_s N_s}{\Delta m_s \rho_r N_r}$$

where Δm_r , Δm_s are the mass losses for reference and tested samples, respectively; ρ_r and ρ_s densities of reference and tested sample, respectively; N_r and N_s are numbers of corresponding rubber roller's cycles, $N_r = N_s = 4490$. These tests have been carried out on 3 reference and 3 coated samples after running-in for 1 h. The test duration of each sample was 5 h. The mass losses have been averaged for all three samples.

Surface roughness has been evaluated using scanning interferential microscope ZYGO NewView 7300.

3. Results and discussion

3.1. Microstructure and phase composition

The microstructures of electron beam deposited composite (Fe-20%Mn-4%V-4%Mo)+10%TiC and (Fe-20%Ni-4%V-4%Mo)+10%TiC coatings are composed of partially dissolved TiC particles in the austenite matrix. On aging 600 °C 1 h the secondary fine carbides precipitate from the austenitic matrix in both types of coatings. According to XRD, these carbides precipitate both from nickel- and

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