Structure and wear resistance of Fe-Cr-Mn-C-N coating obtained by electron beam overlay welding

N A Narkevich¹, V E Panin¹, V G Durakov² and I A Shulepov¹

¹ Institute of Strength Physics and Materials Science SB RAS, 2/4 pr. Akademicheskii, Tomsk, 634055 Russia
² Research and production company «Kubanneftmash», 1 Mekhanizatorov street, Krasnodar region, Akhtyrsy, Russia

³ E-mail: natnark@list.ru

Abstract. The coating produced by electron beam overlay welding of a nitrogen alloyed austenitic steel (24.40Cr, 16.40Mn, 0.18Ni, 1.1Si, 0.57C, 0.70N, and bal Fe (in wt.%) is investigated. Electron beam overlay welding is performed by a continuous low-energy (27 keV), low-current (0.02–0.04 A) focused electron beam using electron beam setup (ISPMS SB RAS, Russia) with a residual pressure of 0.1 Pa. This paper examines microstructure, phase composition, chemical composition of the coating by OM/AES/XRD/SEM/EDS methods. The obtained coating is pore-free and exhibits high work hardening characteristics. Friction loads induce the γ→α′ transformation in the wear process. With increasing friction loads, the friction coefficient decreases.

1. Introduction

Alloys that yield high hardness of 60–65 HRC are widely applicable in industry for wear-resistant overlays. This high hardness is provided by martensitic matrix strengthened with carbides [1, 2]. A substantial disadvantage of these alloys is their crack susceptibility. Coatings obtained by using high-carbon alloys with austenitic matrix do not have the above-mentioned drawback, their wear resistance is greater than for alloys with martensitic matrix [3]. When loaded, including friction loads, (γ) austenite transforms into (α′) deformation martensite [4], which prevents the formation of cracks and wear particles in the wear process. Along with Ni, Mn, and C, nitrogen is a strong austenite-promoting element. Nitrogen austenite is characterized by high strength [5-7], corrosion resistance [8-10], wear resistance [11-14], and a number of other desirable properties. This determines the necessity for the development of methods and techniques for overlay welding of nitrogen alloyed austenitic matrix coatings. Several methods for the production of nitrogen-containing coatings are known including high velocity oxy-fuel spraying process and atmospheric plasma spraying process [15], reactive magnetron sputtering [16,17], plasma sputtering [18], arc surface welding [19], and electron beam overlay welding [20]. Apart from advantageous properties, these coatings exhibit low adhesion strength [16-18] and are not pore-free [19].

In one of our investigations [20], the coating was produced by electron beam overlay welding [21, 22]. This technique is characterized by high heating and cooling rates of 10⁸ - 10¹⁰ K s⁻¹ [23]. High adhesion strength is achieved by mixing in the overlay welding process of the used alloy with the substrate melt to a depth of ≈10 μm. The coating produced by this technique in Ref. [20] did not
contain pores but had heterogeneous phase composition and properties throughout its thickness since a mixture of high-carbon ferrochrome, nitried ferrochrome, ferromanganese, and iron powders was employed for overlay weld. Homogeneous phase composition and structure can be obtained by using nitrogen alloyed austenitic steels, as in Refs. [15-17]. In addition, electron beam overlay welding can be applied to produce multilayer coatings with a thickness of 2-3 mm.

2. Materials and methods
The objective of the present study is to produce and examine the structure and wear resistance of nitrogen alloyed austenitic matrix coatings obtained by electron beam overlay welding of a nitrogen alloyed austenitic steel powder.

The nitrogen alloyed steel powder was produced by the POLEMA JSC company by the method of gas atomization. The size of the powder used for electron beam overlay welding did not exceed 350 μm. The powder’s chemical composition was 24.40Cr, 16.40Mn, 0.18Ni, 1.1Si, 0.57C, 0.70N, and bal Fe (in wt.%). The coating was produced using electron beam setup (ISPMS SB RAS, Russia) [21]. Overlay welding [22] was performed at a residual pressure of 0.1 Pa by a continuous focused electron beam swept in a line with the dimensions of 0.5mm×20mm. The parameters of electron beam overlay welding were as follows: beam current varied in the range I=0.02-0.04 A; maximum electron energy U=27 keV; feed rate of the nitrogen alloyed austenitic steel powder V=0.7 g s⁻¹; rate of the substrate motion relative to electron beam V=3 mm s⁻¹. The coating 20 mm wide and 3 mm thick was welded layer by layer onto a carbon steel substrate increasing its thickness by 0.5 mm with each layer. After overlay welding, the coating was cooled in a vacuum chamber.

Metallographic analysis was carried out using a light microscope (Axiovert-25CA, Zeiss, Germany). Prior to optical metallography (OM), the coating surface was polished up mechanically to a mirror finish in a SAPHIR 550 Grinder/Polisher System (ATMGMBH, Germany) using diamond pastes (3±0.3 μm), and then etched with a mixture of HNO₃ + HCl (1:3 volume ratio). The ratio of various structural elements in optical microstructure was determined by the method of secant planes (Rosival’s method) as the arithmetic mean of 20 measurements [24]. Microhardness of different phases and friction track surface was measured by a NANO Hardness Tester-NHT-S-AX-000X (CSEM, Switzerland) with 10 mN load.

The chemical composition of the coating was examined by Auger electron spectroscopy (AES) (Skhuna-2 Auger spectrometer, Russia) employing 3-keV primary electron beam and 3-keV argon ion beam 1mm in diameter. The chemical composition was defined as the arithmetic mean of 10 spectra. Phase structure and lattice strain were investigated by X-ray diffraction (XRD) using a DRON-4M diffractometer with CoKα radiation and the PDWin program (Bourevestnik, Russia). The phase structure was analyzed in symmetric Bragg-Brentano geometry. Deformation of the face-centered cubic (fcc) lattice was determined through the analysis of the integral breadth of the XRD peaks of the coating after electron beam overlay welding FWHMₘₚf and of a coating fragment after quenching FWHMₘₖf. The volume fraction of the α’ martensite was calculated by ratio according to Ref. [25]. The SEM/EDS analysis of the chemical composition of the matrix and inclusions as well as the morphology of the friction surface was carried out using a Quanta 200 3D system (FEI Company, CIIIA).

Tribological behaviors of the coating were determined in a ball-on-disc wear test using a tribometer (THT-S-AX0000, CSEM, Switzerland) at room temperature. The wear resistance of the high manganese austenitic steel (Hadfield steel) 13.0Mn-1.2C and bal Fe (in wt.%) was studied for comparison. A hard alloy ball (94W-6Co (wt.%)) with a diameter of 3 mm was used against a disc of the investigated material. The wear tests were carried out under a load of 1, 2, 3, 5, 6, 7, 10 N without lubrication at a ball sliding velocity of 30 mm s⁻¹, a trajectory diameter of 4 mm, through 6000 revolutions. The friction coefficient was measured continuously during the tests and recorded on a dedicated data-acquisition computer. The wear resistance was determined as the arithmetic mean of 5 measurements of the cross-section of the friction track using a Micro Measure 3D Station laser profilometer (STIL, France).
3. Results and discussion

3.1. Chemical composition and phase structure

Figure 1 shows the AES spectra that demonstrate changes in the coating’s chemical composition during electron beam overlay welding compared to the original composition of the high interstitial austenitic steel (powder).

According to the AES analysis intensity of peaks of Fe increase after electron beam overlay welding due to evaporation of Cr and Mn. The chemical composition of the coating was investigated by EDS method too. Results are shown in table 1.

|       | Cr  | Mn  | C   | N   | Si  |
|-------|-----|-----|-----|-----|-----|
| AES   | 22.0| 15.8| 0.5 | 0.6 | 1.0 |
| EDS   | 21.2| 12.3| 0.4 | 0.5 | 1.3 |

Figure 2 shows the powder and of the Fe-Cr-Mn-C-N coating diffraction profiles, obtained by electron beam overlay welding. It can be seen that all diffraction profile shows austenite phase. fcc lattice parameter of the coating \(a=0.3618\pm1.1\times10^{-4}\) nm is less than of lattice parameter of powder \(a=0.3623\pm4.0\times10^{-4}\) nm. The XRD pattern in Fig. 2 demonstrates that the family of planes \(200\) in the coating layer adjacent to the substrate has the maximum integral intensity, which corresponds to the crystallization texture \(\{100\}<001\). The \(<001>\) direction coincides with the coating surface normal and with the direction of its crystallization. No texture is observed in the surface layer. The value of the fcc lattice strain in the surface layer appeared to be \(e_{hkl}=3.1\times10^{-4}\), whereas in the coating layer adjacent to the substrate that value was \(e=9.8\times10^{-4}\). The lattice strain in the coating layer adjacent to the substrate with the \(\{100\}<001>\) texture was larger than in the surface layer owing to the fact that the Young’s modulus \(E_{200}\) is lower than other \(E_{hkl}\) [26].

3.2. Structure

Figure 3a demonstrates the OM image of the coating’s structure. The austenite grains are surrounded by darker areas compared to austenite. Grains contain dispersed particles that are not recorded on the XRD pattern owing to their low volume fraction. The volume fraction of the near-boundary regions

![Figure 1. AES spectra of Fe-Cr-Mn-C-N steel (powder) and of the coating produced by electron beam overlay welding.](image1)

![Figure 2. X-ray patterns of the powder and Fe-Cr-Mn-C-N coating, obtained by electron beam overlay welding.](image2)
containing dispersed phases does not exceed 10%. Neither pores nor cracks are observed in the coating. The microstructure of the coating is uniform throughout its thickness. Fig. 3b presents a SEM image of the coating region separating two austenite grains. It can be seen that this region has a complex structure and consists of large irregular-shaped particles and lamellar particles M2N. According to the SEM/EDS analysis the chemical composition of large particles — (Fe, Cr, Mn)71(C, N)29 is close to the M7C3 carbide in which a part of carbon atoms is replaced with nitrogen. Carbonitride particles with a microhardness of 1250 HV precipitate at the grain boundaries. Deformation twins are found in the austenite that surrounds the carbonitrides. These twins are the traces of plastic deformation and indicate high local stress in the austenite structure of the multilayer coating is formed in the conditions of multiple thermal cycling. The number of heating/cooling cycles depends on the number of layers or on the target thickness of the coating. In a 3-mm thick coating, the fraction of the austenite containing carbonitride and nitride particles does not exceed 10%. Consequently, for thinner coatings, the volume fraction of the austenite that does not contain secondary phases will grow.

In austenite grains, there are thin intersecting lines with a dominant orientation in each grain, (Figure 3a). Since in high nitrogen steels (SFE>20 mJ m^-2) the main deformation mechanism is twinning [27-30], it can be assumed that etching has revealed deformation twins and stacking faults in the austenite. Figure 3c shows TEM image of the coating’s structure. Twinning as residual stress relaxation mechanism begins in the coating during its cooling. Assessment of SFE of investigated coating using equation [29]: SFE (mJ m^-2) = -5.97+39.94(C+N)+3.81(C/N) (in wt.%), shows SFE=33-41 mJ m^-2. Twinning-dominant deformation extend to high C+N content up to around 1.0 wt% as well as SFE values (up to 40 mJ m^-2) [30]. This is true for case dissolution of carbides and nitrides at a high
temperature and then high speed cooling (water quenching) preventing reprecipitation. As carbides and nitrides present in the microstructure after overlay welding (Figure 3b), enrichment of austenite by (C+N) decreases. Therefore, precipitation of the carbides and nitrides can decreases SFE of the austenite and increases tendency to \( \alpha' \)-martensite formation.

3.3. Wear resistance
During friction, the (C+N) bearing austenite is continuously hardened. Microhardness of the friction surface increases as the friction load grows (940 HV at 2 N load and 1520 HV at 10 N load). Figure 4 presents a SEM image of the friction track in the coating produced by electron beam overlay welding of the high interstitial austenitic steel powder. The smoothened surface of the friction track indicates plastic deformation of the surface layer by shear without cracking. The XRD pattern reveals the formation of the \( \alpha' \) martensite in the surface layer of the coating’s friction track (Figure 5).

Consequently, friction loads result in the formation of a highly deformed work-hardened surface layer and induce the \( \gamma \rightarrow \alpha' \) transformation. The volume fraction of the \( \alpha' \) martensite formed during deformation by friction is about 30%. The \( \gamma \rightarrow \alpha' \) transformation is not solely a mechanism for material hardening in the frictional contact zone; it is also a stress relaxation mechanism that hinders the formation of wear particles. Table 2 demonstrates the wear resistance and the friction coefficient of the coating and Hadfield steel for comparison depending on normal load.

| Load P, N | Fe-Cr-Mn-C-N coating | Hadfield steel |
|-----------|----------------------|---------------|
|           | \( I, \mu m^2 \text{rev.}^{-1} \) | \( k \) | \( I, \mu m^2 \text{rev.}^{-1} \) | \( k \) |
| 2         | 0.10                 | 0.68          | 0.10 | 0.43 |
| 3         | 0.30                 | 0.70          | 0.15 | 0.45 |
| 5         | 0.27                 | 0.63          | 0.20 | 0.44 |
| 6         | 0.50                 | 0.4           | 0.80 | 0.45 |
| 7         | 0.50                 | 0.4           | 1.35 | 0.48 |
| 10        | 0.85                 | 0.4           | 2.10 | 0.50 |
In the tests with loads $\leq 5$ N, the wear resistance of the coating lower then the wear resistance of Hadfield steel, which, like the investigated coating, is characterized by austenitic structure, high work hardening properties and wear resistance [31]. The microstructure of the coating contain carbonitrides and nitrides. Hence, the particles/matrix interfaces associated with high stress concentration. Therefore existence of particles in coating’s structure leads to decohesion at the particles/matrix interfaces during the frictional contact. With increasing friction load more then 5 N decohesion continues, but work hardening of the high interstitial austenitic coating became higher then that of one phase austenitic Hadfield steel. With increasing friction loads more then 5 N wear resistance appears to be 1.5-2 times greater than that of Hadfield steel as well as the friction coefficient $k$ is reduced to 0.4.

4. Conclusions

The multilayer coating produced by electron beam overlay welding of nitrogen alloyed austenitic steels is pore-free. The phase composition of the coating is uniform and consists of the austenitic matrix, $\text{Me}-(\text{C, N})$: carbonitrides, and $\text{Me}_2\text{N}$ nitrides. The coating does not deteriorate under the action of residual stress. The presence of twins in the structure indicate plastic deformation induced by residual stress. Therefore, twinning in the coating represents deformation mechanism, mechanism work-hardening and residual stress relaxation mechanism, preventing the formation of cracks after electron beam overlay welding.

The austenite is characterized by high work hardening. The increasing load during friction leads to the $\gamma \rightarrow \alpha'$ transformation that causes a reduction in friction coefficient.

The composition of the powder for surfacing can be modified by adding refractory carbide compounds, for example titanium carbides, vanadium carbides. When using a matrix in the form of nitrogen-containing austenite hardened by carbides, we predict the formation of a corrosion-resistant composite coating, resistant to cracking, with high work hardening and increased wear resistance.

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