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Comparative study between high-velocity oxygen fuel and flame spraying using MCrAlY coats on a 304 stainless steel substrate

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A B S T R A C T

High-velocity oxygen fuel (HVOF) and flame spraying (FS) are alternative methods to thermal spraying processes to produce dense high-quality coatings. The aim of this study is to compare the microstructure and mechanical properties for HVOF and FS using a thermally sprayed bond coat NiCoCrAlY and CoNiCrAlY powders on an AISI 304 stainless steel substrate. The microstructure and composition of coatings were characterized by an X-ray Diffraction (XRD) analysis and Electron Microscopy (SEM) coupled to an Energy Dispersive Spectroscopy (EDS) detector. HVOF showed higher quality coatings compared to FS in terms of porosity and the presence of unfused particles for both employed powders. The mechanical properties and results indicated that the yield strength of the NiCoCrAlY (HVOF) coating was 1.4-folds FS, but the flexural bending modulus was almost the same. For the CoNiCrAlY powder, HVOF gave a higher yield strength and a higher flexural modulus than FS as the oxygen affinity of the CoNiCrAlY powder was lower than NiCoCrAlY given the high Co content in the former versus the latter. The results also indicated that hardness increased by about 83%, and by 58% for the NiCoCrAlY HVOF and FS coating alloys, and by 42% and 20% more for the CoNiCrAlY HVOF and FS coatings than the stainless steel substrate hardness, respectively.

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1. Introduction

Thermally sprayed coatings are normally used to enhance surface properties of nickel super alloys. These alloys are of widespread use for aerospace and gas turbine fields. MCrAlY powders are employed as thermal insulation coatings (mainly
yttria stabilized partially with zirconia). Today’s coating technology is improving, but the success of a new technology depends on coating performance, cost and environmental impact [1]. Hence the use of different techniques for depositing similar coatings may lead to various results in quality and cost terms. Thermal spray coatings are frequently used as oxidation resistant materials (mainly MCrAlY alloys where M stands for Co, Ni, or for both) and thermal insulation coatings (mainly yttria stabilized partially with zirconia, YSZ). The standards of the original equipment manufacturers (OEMs) recommend MCrAlY alloys being deposited using HVOF. However, other techniques, such as flame spraying (FS), are still recommended given their lower cost. Some researchers insist that the quality of MCrAlY coatings deposited by HVOF is lower due to the partial oxidation that takes place during spraying [2]. Nevertheless, this lower quality is rarely quantified according to experimental evidence. It has been observed that highest quality coats are obtained by laser cladding [3], but this is very expensive compared to HVOF and FS, whose cost is much lower than laser cladding. The aim of this study is to compare the microstructure (porosity, oxide concentrations and presence of unmelted particles) and mechanical properties (microhardness and three bending tests) of HVOF and FS using thermally sprayed bond coat NiCoCrAlY and CoNiCrAlY powders on an AISI 304 stainless steel substrate.

2. Materials and experimental procedure

2.1. Material and method

Commercially nickel-base NiCoCrAlY (Ni 23Co 17Cr 12Al 0.5Y wt%) from Sulzer Metco Amdry 365-2 (55 ± 17 μm), and cobalt-base CoNiCrAlY (Co 32Ni 21Cr 8Al 0.5Y wt%) from Sulzer Metco Amdry 995C (64 ± 20 μm) powders were used to spray them on austenitic stainless steel AISI 304 on samples (200 × 30 × 3 mm) by HVOF and the Oxygen Fuel (OF) spraying system, model DJ-2600/2700 (Sulzer Metco Inc., Westbury, NY, USA). An ultrasonic bath was used to clean the surface with acetone. The average surface roughness was 5.1 ± 0.5 μm and the mean roughness depth was 28.3 ± 2.4 μm. Coatings were deposited by a spray gun CDS-80000 (Rutectic Castolin mark) with particles velocity at around 300 m/s and a peak temperature of 3200 °C. The projection parameters of FS are tabulated in Table 1. The HVOF process parameters employed for this research are shown in Table 2.

<p>| Table 1 – Parameters of the FS process. |</p>
<table>
<thead>
<tr>
<th>Sample</th>
<th>Preheating strokes</th>
<th>Standoff distance (mm)</th>
<th>MCrAlY coat strokes</th>
<th>Coat thickness (average) (μm)</th>
<th>Comp. air pressure bar</th>
<th>Flame condition</th>
</tr>
</thead>
<tbody>
<tr>
<td>AISI 304</td>
<td>1.2</td>
<td>150</td>
<td>9</td>
<td>186.5</td>
<td>3</td>
<td>Neutral</td>
</tr>
</tbody>
</table>

<p>| Table 2 – Thermal spraying parameters applied on HVOF coatings. |</p>
<table>
<thead>
<tr>
<th>HVOF coating</th>
<th>Oxygen</th>
<th>Propylene</th>
<th>Air</th>
<th>Powder feed rate (g/min)</th>
<th>Standoff distance (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>AISI 304</td>
<td>Pressure (bar) Flow rate (SLPM)</td>
<td>Pressure (bar) Flow rate (SLPM)</td>
<td>Pressure (bar) Flow rate (SLPM)</td>
<td>7.2384</td>
<td>38</td>
</tr>
<tr>
<td></td>
<td>10.3</td>
<td>139</td>
<td>6.9</td>
<td>89</td>
<td></td>
</tr>
</tbody>
</table>

2.2. Characterization of MCrAlY coatings

Coatings’ cross-sections were metallo-graphically prepared and samples’ morphology was observed with a Hitachi SEM (TM3000, Tokyo, Japan) attached to an EDS unit. The phases of the top and under coats were characterized by XRD in an X’Pert PRO diffractometer (Kristallogrex 5000, Siemens, Germany) with Cu Kα (λ = 0.154 nm) radiation and a range of diffraction angles (2θ) between 20° and 120°.

2.2.1. Porosity and microhardness

Porosity was characterized by an image analyzer with the Image J software [3]. This software converts coatings’ micrographs into binary images, and quantifies the percentages of the two colors. Microhardness measurements were taken on the coatings’ polished cross-sections according to Standard ASTM E384 with a Vickers indenter (Shimadzu, HMV-2, Tokyo, Japan) using a 0.3 kg load and a 15 s dwell time [3-10]. Each coating’s microhardness was determined as the average of 20 measurements taken on three different samples.

2.2.2. Three-point bending test measurements

The three-point bending test of the coated samples was carried out in the universal test machine (Shimadzu model AG-X, Japan) with a 50 kN load-cell. Camcorder extensometry was used. The backside of the substrate was at the top, touched by the hardened steel punch, while the coating was at the bottom, as shown in Fig. 1. The linear displacement loading pattern was imposed onto the coating samples and load was able to achieve a load-cell up to 50 kN, while the loading rate

![Fig. 1 – Schematic of the three-point bending test configuration.](image-url)
was 0.05 mm/min with constant displacement control. The deflection on the outer coating surface was measured during the test by an Epsilon Technology digital deflectometer, model 3540-004 M-ST. The diagram of force versus displacement was displayed in relation to time during the three-point bending test. When the coating separated from the substrate and the bending moment reached a certain value, the bending test was stopped and no substrate failure took place. The bending nominal stress (σ bending) and the strain (ε ext.) of the outer (coating surface) were calculated using Eqs. (1) and (2), respectively.

\[
\sigma_{bending} = \frac{3L.F}{2b.t^2} \tag{1}
\]

\[
\varepsilon_{ext} = \frac{6.t.d}{L^2} \tag{2}
\]

where \(L\) is the distance between the lower supports; \(F\) is the applied centered load; \(b\) is the sample width; \(t\) is the total sample thickness (substrate/coating) and \(d\) is the deflection measured at the centerline of the outer coating surface by the digital deflectometer.

3. Results and discussion

The results of microstructure and mechanical properties for the HVOF and FS techniques using thermally sprayed bond coat NiCoCrAlY and CoNiCrAlY powders on an AISI 304 stainless steel substrate are presented and discussed in the next section.

3.1. Geometry, porosity and microstructure of coatings

The optical microstructures of the as-sprayed NiCoCrAlY and CoNiCrAlY coatings using two different coating types are shown in Fig. 2. The top-coats mean thicknesses were measured by the optical microstructures of the cross-sections. The coatings’ thickness and porosity values are offered in Fig. 2. The CoNiCrAlY alloy coating thicknesses were 360 ± 25.47 (FS) and 388 ± 21.87 μm (HVOF) (Fig. 2a and b). The initial coating thickness of the NiCoCrAlY alloy was 300 ± 22.87 (FS) and 325 ± 22.94 μm (HVOF) (Fig. 2c and d). The quantitative evaluation of porosity was made by the image analyzer software. Porosity of the CoNiCrAlY alloy was 16.65 ± 5.57% (Flame Spraying), 14.62 ± 2.87% (HVOF) and the NiCoCrAlY alloy coating was 25.34 ± 6.90% (Flame Spraying) and 23.36 ± 2.37% (HVOF), respectively.

The BSE–SEM micrographs of the as-sprayed NiCoCrAlY and CoNiCrAlY coatings both displayed a similar structure for the two different coating types presented in Fig. 3. However, the results revealed that the number of unfused particles of the FS coating was significantly larger than for the HVOF coatings, as shown in Fig. 3a–d. The content of unfused particles was calculated by the quantitative image analysis. The NiCoCrAlY and CoNiCrAlY coatings showed a homogeneous microstructure and an adhesive interface between the coat and substrate, but without visible connected pores. The NiCoCrAlY coatings (Fig. 3a–d) exhibited higher levels of porosity and clearer interlamellar boundaries than the CoNiCrAlY coatings. They also show a higher proportion of larger pores and semimelted particles (this semimelted pro-
portions in the CoNiCrAlY coating were higher than in the NiCoCrAlY coating. The different contrast regions in the NiCoCrAlY micrograph in Fig. 3d correspond to different phases: β-NiAl and γ-NiCoCr [12–16].

Fig. 4 shows the SEM micrograph and elemental mapping of a cross-section of the as-sprayed (a) NiCoCrAlY (HVOF), (b) CoNiCrAlY (HVOF), (c) NiCoCrAlY (FS) and (d) CoNiCrAlY (FS) coatings. The Al particles are homogeneously distributed on both CoNiCrAlY and NiCoCrAlY by different coatings; most of the Cr element and part of the Co element are independently distributed, which suggests lots of metallic Cr and Co on this MCrAlY coat, which is consistent with the related XRD results. Fig. 4 shows the morphology of the cross-section and corresponding element mapping of the MCrAlY coatings. The scale found on the surface was still α-Al2O3, while the β-NiAl phase had disappeared. The EDS mappings data agree with the data obtained by XRD superficial analysis, which showed that the scale was formed by NiO and by protective Cr2O3 metallic phases. The Al2O3 phase surrounding the Cr-Ni metallic splats was responsible for improving the coat’s oxidation resistance because Al2O3 formation inhibited the diffusion of Ni and Cr from inside the splat to the surface and, thus, hindered the growth of Cr2O3 and NiCr2O4. This may be attributed to the low diffusion coefficients of Ni and Cr along the grain boundary of the metallic phase. The diffusion coefficient of aluminum along the grain boundary of Ni and Cr was higher than that of Ni and Cr [19]. The above discussion evidenced that the efficiency of a thermally sprayed NiCrAlY coating on austenitic stainless steel was good owing to the formation of a continuous Cr2O3 and Al2O3 double layer beside the coating’s internal oxidation [19]. To analyze the distribution of the chemical elements in dust, EDS studies were carried on particles. Fig. 4a–d shows a characteristic spectrum that reveals the existence of chemical elements in the alloy and the relative percentage. A correlation was found between the composition reported by the supplier of the alloy and the experimental results of the tests herein performed. Elemental mapping was also performed on dust particles to determine the distribution of Co, Ni, Cr, Al, and Y in them. Fig. 4 shows the results of this mapping in a dust particle, which reveals that the alloying elements are distributed homogeneously and according to the chemical composition. This means that aluminum and yttrium have a lower content, but a uniform distribution. No segregation toward certain areas of the studied particles was noted. This behavior was observed in both the elongated and rounded particles, and no change was noted in the elemental distribution geometry.

3.2. The XRD analysis of the MCoCrAlY coatings

All the coatings retained the γ and β phases consistently with the first set of experiments, and no presence of oxides and/or secondary phases was noted in the XRD patterns. The XRD patterns obtained from the surface of the FS and HVOF thermally sprayed NiCoCrAlY and CoNiCrAlY samples (Fig. 5) evidenced the existence of stable Cr2O3 and NiO phases and the CrNi metallic phase.

3.3. Microhardness analysis of the MCrAlY coatings

The evolution of hardness from the coatings surface of the MCrAlY coatings to the substrate by different coating meth-
ods (HVOF and FS) is shown in Fig. 6 and the results are summarized in Table 3. Coatings’ hardness is greater than that of the stainless steel substrate, and the hardness of the NiCoCrAlY coatings was greater than the CoNiCrAlY coatings in both coating methods. The results also indicated that hardness increased by about 83 and by 58% for the NiCoCrAlY HVOF and FS coating alloys, and by 42% and 20% more for the CoNiCrAlY HVOF and FS coatings than the stainless steel substrate hardness, respectively. This difference in hardness behavior between coatings may be due to the coating microstructure and the formed phases, which more β phase stabilization found in the coating with a higher aluminum content (12.5 wt%) (NiCoCrAlY) than in the CoNiCrAlY alloy (8 wt%), which is similar to Juan et al. [7]. The average microhardness value was approximately 323 ± 27 HV0.01 (HVOF), whereas values of about 277 ± 15 HV0.01 (FS) were measured in pore-free areas (denser sites). The value in the porous areas was 140 ± 30 HV0.01, which was due to the poorer coating structure in these areas [20].

3.4. Bending strength of the MCrAlY coatings

The plot of the bending stress versus the bending strain for the two different types of coating methods (HVOF and FS) is
**Fig. 5** – XRD analysis (a) NiCoCrAlY (HVOF) and FS coating, (b) CoNiCrAlY (HVOF) and FS coating.

### Table 3 – Comparison table of the average microhardness results for the MCrAlY alloy coatings obtained by HVOF, FS, thermal spray processes and substrates.

<table>
<thead>
<tr>
<th>Coating/substrate material</th>
<th>Process</th>
<th>Microhardness (kg/mm²)</th>
<th>HV scale</th>
<th>Reference</th>
</tr>
</thead>
<tbody>
<tr>
<td>NiCoCrAlY</td>
<td>HVOF</td>
<td>321 ± 16</td>
<td>HV0.01</td>
<td>Own measurements</td>
</tr>
<tr>
<td></td>
<td>FS</td>
<td>277 ± 21</td>
<td>HV0.01</td>
<td>Own measurements</td>
</tr>
<tr>
<td></td>
<td>HVOF</td>
<td>434 ± 64</td>
<td>HV0.01</td>
<td>Mercier et al. [24]</td>
</tr>
<tr>
<td></td>
<td>Plasma spray</td>
<td>450 ± 45</td>
<td>HV0.01</td>
<td>Brodin et al. [21]</td>
</tr>
<tr>
<td></td>
<td>Coaxial laser cladding</td>
<td>492 ± 13</td>
<td>HV0.01</td>
<td>Juan et al. [20]</td>
</tr>
<tr>
<td>CoNiCrAlY</td>
<td>HVOF</td>
<td>248 ± 13</td>
<td>HV0.01</td>
<td>Own measurements</td>
</tr>
<tr>
<td></td>
<td>FS</td>
<td>203 ± 19</td>
<td>HV0.01</td>
<td>Own measurements</td>
</tr>
<tr>
<td></td>
<td>Vacuum plasma spray</td>
<td>398 ± 37</td>
<td>HV0.03</td>
<td>Scrivani et al. [22]</td>
</tr>
<tr>
<td></td>
<td>HVOF</td>
<td>410 ± 35</td>
<td>HV0.03</td>
<td>Scrivani et al. [22]</td>
</tr>
<tr>
<td></td>
<td>Axial plasma</td>
<td>391 ± 40</td>
<td>HV0.03</td>
<td>Scrivani et al. [22]</td>
</tr>
<tr>
<td></td>
<td>Plasma spray</td>
<td>155 ± 18</td>
<td>HV0.03</td>
<td>Higuera et al. [23]</td>
</tr>
<tr>
<td></td>
<td>Coaxial laser cladding</td>
<td>361 ± 15</td>
<td>HV0.01</td>
<td>Juan et al. [20]</td>
</tr>
<tr>
<td>Substrate AISI 304</td>
<td>–</td>
<td>175 ± 6</td>
<td>HV0.01</td>
<td>Own measurements</td>
</tr>
</tbody>
</table>

### Table 4 – Three-point bending tests of the mechanical behavior of the MCrAlY coating systems results.

<table>
<thead>
<tr>
<th>Coating/substrate material</th>
<th>Process</th>
<th>Bending modulus (GPa)</th>
<th>Bending yield strength (MPa)</th>
<th>Failure strength (MPa)</th>
<th>Deformation at failure (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>NiCoCrAlY</td>
<td>HVOF</td>
<td>103.46 ± 5.87</td>
<td>542.72 ± 22.13</td>
<td>822.82 ± 20.06</td>
<td>1.15 ± 0.12</td>
</tr>
<tr>
<td></td>
<td>FS</td>
<td>101.54 ± 4.77</td>
<td>479.72 ± 21.51</td>
<td>857.30 ± 77</td>
<td>0.55 ± 0.15</td>
</tr>
<tr>
<td></td>
<td>Coaxial laser cladding (longitudinal) [20]</td>
<td>118.29 ± 11.66</td>
<td>702.24 ± 98.81</td>
<td>902.59 ± 70.84</td>
<td>1.24 ± 0.28</td>
</tr>
<tr>
<td></td>
<td>Coaxial laser cladding (cross-section) [20]</td>
<td>123.19 ± 4.41</td>
<td>612.30 ± 91.57</td>
<td>766.60 ± 99.15</td>
<td>0.92 ± 0.27</td>
</tr>
<tr>
<td>CoNiCrAlY</td>
<td>HVOF</td>
<td>68.68 ± 15.87</td>
<td>602.01 ± 31.16</td>
<td>805.19 ± 76.60</td>
<td>2.54 ± 0.39</td>
</tr>
<tr>
<td></td>
<td>FS</td>
<td>65.59 ± 17.61</td>
<td>412.41 ± 14.52</td>
<td>626.41 ± 106.9</td>
<td>0.65 ± 0.18</td>
</tr>
<tr>
<td></td>
<td>Coaxial laser cladding (longitudinal) [20]</td>
<td>125.46 ± 7.94</td>
<td>731.67 ± 47.61</td>
<td>1029.61 ± 6.97</td>
<td>3.41 ± 0.53</td>
</tr>
<tr>
<td></td>
<td>Coaxial laser cladding (cross-section) [20]</td>
<td>136.30 ± 13.15</td>
<td>730.33 ± 54.28</td>
<td>840.85 ± 27.87</td>
<td>1.51 ± 0.42</td>
</tr>
<tr>
<td>Substrate</td>
<td>–</td>
<td>146.31 ± 4.11</td>
<td>546.83 ± 50.16</td>
<td>–</td>
<td>–</td>
</tr>
</tbody>
</table>
shown in Fig. 7a and d, respectively. The summarized results are shown in Fig. 7c and d, respectively and Table 4. It is clear that the plot of the bending stress versus the bending strain for the substrate and the MCrAlY coating was nearly the same for the HVOF and CoNiCrAlY coatings. This was due to the bond-coat being a metallic layer, the thermal expansion coefficient of the bond-coat came close to that of the ceramic top-coat and the substrate, and the bonding strength between the substrate and bond-coat was usually very high compared to the bonding strength between the bond-coat and top-coat [16,18,20]. The bending strength for the FS coating was nearly the same with the substrate, but the CoNiCrAlY coating decreased for FS because of the weakness of bonding for this coating with the substrate, as shown in the microstructure analysis. When the applied load was not too high, the bond-coat and substrate deformed together. However, when the ceramic top-coat was deposited on the surface of the bond-coat, the curve deviated in the initial plastic deformation stage. The corresponding flexure stress increased with the same flexure strain compared to the superalloy/bond-coating system, which may be attributed to the higher young’s modulus of the ceramic top-coat compared to the young’s modulus of the metallic substrate. Plastic deformation needs more stress and energy. The stainless steel substrate usually has lower elastic modulus and hardness values compared to the super alloy substrate, thus it needs a lower stress level when the same strain for the stainless steel/MCrAlY coating system takes place.

3.5. Fractography of the MCoCrAlY coatings

Fig. 8 depicts the cross-section for the coating after the bending test. Fig. 8a shows the image of the cross-sectional image for the as-sprayed coating, and depicts how TBCs have double layers; i.e., bond-coat and top-coat, and the thickness of the bond-coat and top-coat is about 120 μm and 300 μm, respectively. The pore sand microcracks are distributed randomly over the inside of the ceramic layer. The bond-coat/top-coat interface exhibits the characteristic fluctuation. When the coating was endured with a three-point bending (3PB) test, the large crack accompanied by a large gap at the center of the cross-section is obtained, which may be due to the large loading force applied to the center line of the coated samples. With an increasing distance away from the central part, the crack becomes smaller, and the distribution and characteristic cracks seem to be axisymmetric with the center line taken as the symmetry axis. The crack becomes smaller the further away it is from the central part. When the distance between the position and the central part is long enough, no cracks initiate with a certain applied load (see Fig. 9). The crack initiates on the coating’s surface and propagates toward the top-coat/bond-coat interface after passing through the interface before finally reaching the substrate/bond-coat interface. The horizontal crack initiates at the substrate/bond-coat interface.

Fig. 6 – Hardness of the MCrAlY alloy coatings/substrate HVOF and FS.

Fig. 7 – The bending stress–strain curves for the different types of coating methods of NiCoCrAlY and CoNiCrAlY on substrate AISI 304 (a) HVOF, (b) FS and mechanical properties clear in the (c) HVOF and (d) FS coatings.
interface and propagates in the substrate/bond-coat interface direction. Then the coating fails eventually at the central part of the substrate/bond-coat interface for the coated samples, which indicates cracks toward the MCrAlY/substrate interface. From the cross-section of the MCrAlY/substrate specimen (Fig. 9a), macrocracks penetrate perpendicularly over the whole overlay coating and branch symmetrically along the MCrAlY/substrate interface, before being deflected onto the substrate. A small-sized crack is also found near the coating surface. Therefore it is assumed that crack nucleation may start from the coating surface. In the HVOF coating-exposed specimen, the longitudinal zigzag crack is observed to extend through the NiCrAlY coating before deflecting on the substrate (Fig. 9b). A careful observation shows that flaws emanate from the defects on the NiCrAlY interface. The distortion zone in the fracture morphology (Fig. 9c) proves that the NiCrAlY coating suffered a fatigue fracture, which may be due to the load fluctuation during the test. The NiCrAlY coating shows large dimples, which means that the coating becomes ductile after thermal exposure. The stress change is the result of thermal and growth stresses. During thermal exposure, the replacement reaction of $\text{Cr}_2\text{O}_3$ with $\text{Al}_2\text{O}_3$ was run with a lower density on the barrier layer, and defects formed close to the layer with a difference in the CTEs of the oxides from the alloy substrate or the over layer altering residual stress. The presence of residual stress promoted crack propagation along the interface and resulted in early interfacial damage. The stronger the interfacial stress, the easier crack propagation became along the interface [16]. The reason for the rapid damage of the annealed MCrAlY/substrate specimen would stem from the combination of interfacial residual stress and interfacial products.
Fig. 9 – The SEM–BSE fracture micrographs and macroimage of the MCrAlY FS coating on stainless steel AISI 304 after the three-point bending test (a, c and e) the NiCoCrAlY and (b, e, f) CoNiCrAlY alloy. Arrows are used to indicate the cracks on the MCrAlY/substrate interface.

4. **Conclusions**

Dense MCrAlY coatings, with suitable demonstrated metallurgical bonding with the substrate, were obtained by the HVOF and FS coatings by a bending test, which were comparatively studied in this work. Apart from confirming the mechanical properties and damage behaviors of the thermally sprayed coatings in microstructural details, according to the results obtained in this study, the following conclusions can be drawn:

1. The HVOF coatings were composed of a lamellar structure and unmelted particles, while the as-sprayed FS coating consisted mainly of unmelted particles. The β phase content increased with the presence of unmelted particles. The sprayed HVOF coatings had similar microstructures, while the FS coatings exhibited higher porosity and oxide content. The unmelted particles increased bond coat roughness, which resulted in a localized stress concentration at the top-coat/bond coat interface by many unmelted particles, which may accelerate both the initiation and propagation of cracks on the top-coat.

2. The bending strength of a CoNiCrAlY (FS) coating was poor. On the contrary, it was high for a CoNiCrAlY (HVOF) coating. Yield strength significantly increased for the NiCoCrAlY (HVOF) coating. It was found that a NiCoCrAlY (HVOF and FS) coating with TBC stainless steel had a higher strength than for a CoNiCrAlY (HVOF and FS) coating and a bared stainless steel.

3. The microhardness of the FS and HVOF MCrAlY coatings had a greater microhardness than the AISI 304 stainless steel substrate. The NiCoCrAlY (HVOF) coating was 1.5-fold greater than the FS coating. The high particles velocity during coating was effective in improving adhesive strength. The results also indicated that hardness increased by about 83%, and by 58% for the NiCoCrAlY HVOF and FS
coating alloys, and by 42% and 20% more for the CoNiCrAlY HVOF and FS coatings than the stainless steel substrate hardness, respectively. The aluminum-rich β-phase was harder than the γ-phase, and also in the MCrAlY HVOF coatings with the γ/β microstructure.

(4) Different coated samples exhibited various damage behaviors during the bending test. Different damage behaviors were attributed to the distinct interfacial strength and mechanical properties of MCrAlY. The NiCoCrAlY (FS) coating exhibited poor interfacial strength, which presented an elastic-fragile behavior, with a failure mechanism by delamination and cracks from the interface with the substrate. The NiCoCrAlY (HVOF) and FS coatings possessed greater interfacial strengths, which presented an elastic–plastic behavior, crack failure with no delamination and high ductility.

(5) The bending moduli of coatings NiCoCrAlY and CoNiCrAlY were lower than the substrate; this property has an inversely proportional behavior to the hardness of the HVOF and FS coatings; thus, the bigger the β-(Ni,Co)Al phase, the lower the bending modulus and the less ductility obtained in the HVOF coatings with the γ/β microstructure.

These results confirmed that the potential impact of the thermal FS process is a suitable tool to industrially produce NiCoCrAlY bond coats at a lower cost than the HVOF thermal spraying systems. Both the HVOF and thermal FS bond MCrAlY coats are promising, particularly those coatings with coarser particles, in terms of porosity level and oxygen content, to be used for TBC applications. They come over as a promising application for future comparative studies that employ laser metal deposition (LMD) and laser cladding (LC) with MCrAlY alloys.

Conflicts of interest

The authors declare no conflicts of interest.

Uncited references

[11,17].

REFERENCES


[17] Richer P, Yandouzi M, Beauvais L, Jodoin B. Oxidation behaviour of CoNiCrAlY bond coats produced by plasma,


