Study on production of modified MCrAIY powder with nano oxide dispersoids as HVOF thermal spray feedstock using mechanical milling

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Keywords: MCrAlY coatings, NiCoCrAlY, HVOF thermal spray, nanocomposite coating, ball milling, mechanical milling

Abstract
This paper investigates NiCoCrAlY-CeO2 nanocomposite powders prepared by mechanical milling process. At first, micron-sized ceria powder was pre-milled to obtain nano-sized powders and then 1 wt% CeO2 was mixed with the conventional NiCoCrAlY powder and milled under three sets of parameters. The conventional and milled powders were deposited on low carbon steel substrates by HVOF thermal spray technique. X-ray diffraction (XRD), scanning electron microscopy (SEM), field-emission scanning electron microscopy (FESEM) and image analyzing method were used to characterize the prepared powders and coatings. Microstructural characterization showed that the milled powders underwent morphological and phase transitions. Through the microstructural and compositional analyses, it was found that a homogeneous distribution of CeO2 nano-dispersoids in the NiCoCrAlY matrix was obtained. The resulting coatings showed that by using optimized milling parameters, required particle size and suitable morphology for thermal spray can be achieved.

1. Introduction

The hot parts of gas turbines are generally made of nickel-based superalloys owing to their excellent mechanical properties at high temperatures [1]. These alloys are designed to maintain good creep strength and to withstand high mechanical loads [2]. However, the development of these alloys with superior mechanical properties could be achieved at the cost of detrimental high temperature oxidation and hot corrosion resistance. With regard to oxidation properties, these superalloys are prone to high temperature corrosion and must be protected by applying coatings [3]. Advancements of high-temperature coatings such as diffusion coatings (aluminides or Pt-aluminides), MCrAlY (M = Ni, Co or Ni+Co) overlay coatings and thermal barrier coatings (TBCs) have aided significantly to protect hot parts of turbines during service [4, 5].

MCrAlY coatings have widely been used in gas turbine engines and the oxidation protection of these coatings is provided by developing a thermodynamically stable with high structural integrity, slow growing and adherent α-Al2O3 scale over the coating surface as a result of coating oxidation during high temperature exposure [6]. So, an oxide scale can be formed on the coating, often referred to as a thermally grown oxide (TGO) layer which grows in either parabolic or sub-parabolic manner [7]. In fact, the composition and growth rate of the TGO layer varies according to the coating properties such as the chemical composition, microstructure, and oxides content [8]. These coatings have been manufactured by methods of physical vapor deposition (PVD) [9], magnetron sputtering [10], arc ion plating (AIP) [11], thermal spraying [12–15] and laser powder deposition (LPD) [16]. The MCrAlY coatings deposited using the HVOF process often have better properties in comparison to the other thermal spraying methods and in some cases provided higher oxidation resistance [17, 18]. Also, the HVOF sprayed coatings are able to maintain nano-scale structure after spraying [19]. Moreover, low flame temperature, high particle velocities, small flame-particle interaction time and low temperature rise of the substrate can be mentioned as the advantages of the HVOF process which provides dense coatings with low oxygen content levels and low porosity which exhibits good adhesion to the substrate [20, 21].
Several attempts have been proposed and tried to improve the high-temperature oxidation performance of the MCrAlY coatings [22–25]. In recent years, the nanostructured MCrAIY coatings have attracted increasing research attention due to their ability to form a dense protective, uniform and thin Al₂O₃ external layer on the coating during oxidation [4, 19, 26–30]. The improved oxidation behavior of these coatings can be attributed to the grain size of the deposited coating which plays a key role in the TGO feature. Coatings with nanosized grains could lead to fast diffusion paths for Al from coating through the grain boundaries which favors to the formation of pure and compact protective oxide layer of α-Al₂O₃ [19]. In this regard, mechanical milling process is known to be one of the effective techniques and rapid processing route for preparing nanostructured and nanocomposite powders for the thermal spray deposition processes [26, 31–35]. At present, there are some researches on the effect of several reinforcements in the MCrAlY-based nanocomposites, investigating the mechanical properties and high temperature corrosion behavior of these materials [25, 36, 37].

The objective of the present study is to investigate the effect of mechanical milling to prepare a modified NiCoCrAlY powder with nanocrystalline grains consisting nano ceria particles. The microstructure, morphology and particle size distribution of powders were also investigated. The powders were sprayed using the HVOF thermal spray technique. A detailed discussion of the structural characteristics and high temperature oxidation behavior of these coatings was given in our previous work [38].

2. Experimental procedure

2.1. Preparation of powders

Preparation of samples involves two steps. In the first stage, the thermal spraying powders was prepared while in the second one, these powders were deposited to prepare the coatings with thickness in the range of 200–220 μm. Conventional pre-alloyed NiCoCrAlY powder (AMDRY 365-1) was used as starting spray material. CeO₂ powder was used as reinforcement after milling for 48h (figure 1). In this regard, commercial ceria powder with 99.9% purity and sub-3 micron-sized particles was pre-milled at a revolution speed of 400 rpm and the ball to powders weight ratio was 20:1. Then a mixture of NiCoCrAlY-1wt% CeO₂ was mechanically milled under three sets of milling parameters. The milling was performed by a planetary ball mill using AISI 52100 steel balls as the milling media.

In order to prevent oxidation of the powder blend, the milling vial was filled with high purity argon gas and the milling operation was carried out at room temperature. Also, the experiments were carried out by alternating 60 min of milling with 15 min of rest to inhibit excessive overheating of the vials. For powder characterization, a
small amount of the milled powders was taken out of the vial at specified intervals. Table 1 shows the milling conditions and coding of the samples.

### Table 1. Coding NiCoCrAlY-CeO₂ powders with different milling conditions.

| Sample number | Revolution speed (RPM) | Ball to powder ratio (wt%) | Process control agent (PCA) | Milling time (h) |
|---------------|------------------------|---------------------------|----------------------------|-----------------|
| Conventional  | —                      | —                         | —                          | —               |
| A6            | 200                    | 10:1                      | Ethanol                    | 6               |
| A12           | 200                    | 10:1                      | Ethanol                    | 12              |
| A18           | 200                    | 10:1                      | Ethanol                    | 18              |
| A24           | 200                    | 10:1                      | Ethanol                    | 24              |
| A36           | 200                    | 10:1                      | Ethanol                    | 36              |
| A48           | 200                    | 10:1                      | Ethanol                    | 48              |
| B3            | 350                    | 20:1                      | Ethanol                    | 3               |
| B6            | 350                    | 20:1                      | Ethanol                    | 6               |
| B9            | 350                    | 20:1                      | Ethanol                    | 9               |
| B12           | 350                    | 20:1                      | Ethanol                    | 12              |
| B15           | 350                    | 20:1                      | Ethanol                    | 15              |
| B18           | 350                    | 20:1                      | Ethanol                    | 18              |
| C6            | 200                    | 10:1                      | Stearic Acid               | 6               |
| C12           | 200                    | 10:1                      | Stearic Acid               | 12              |
| C18           | 200                    | 10:1                      | Stearic Acid               | 18              |
| C24           | 200                    | 10:1                      | Stearic Acid               | 24              |
| C36           | 200                    | 10:1                      | Stearic Acid               | 36              |
| C48           | 200                    | 10:1                      | Stearic Acid               | 48              |

### Table 2. HVOF spraying parameters for conventional and milled powders.

| Parameters                  | Conventional | B12  | C24  |
|-----------------------------|--------------|------|------|
| Propane flow rate (l/min)   | 55           | 55   | 55   |
| Oxygen flow rate (l/min)    | 210          | 200  | 225  |
| Powder feed rate (rpm)      | 430          | 400  | 450  |
| Carrier gas (N₂) flow rate (l/min) | 20      | 20   | 20   |
| Spray distance (mm)         | 280          | 250  | 250  |

2.2. Coating deposition

Powders were dried in an oven at 60 °C for 1 h and then sprayed onto the steel substrates to form a coating, using high velocity oxygen fuel (HVOF) spraying technique. The spraying parameters are listed in table 2 and due to the differences in structure and morphology between the three types of the starting powders, the spraying parameters are not the same. The substrate was cooled with compressed air jets during and after spraying.

2.3. Characterization techniques

The morphological and microstructural characterization of the powder particles and coatings were examined by scanning electron microscopy (SEM) and field-emission scanning electron microscopy (FESEM) equipped with Energy Dispersive Spectroscopy (EDS). X-ray diffraction (XRD) analysis was carried out with Cu Kα radiation (λ = 1.54056 Å) with a step size of 0.026° to determine the phases of the powders and coatings. The operating voltage and current were 40kV and 40mA, respectively. Also, the amount of the oxide and porosity in the specimens and the powders agglomerate size distribution were estimated using image J analysis software (NIH, USA). At least 10 micrographs with magnification of 1000 × on the polished cross-section of as-sprayed coatings were used for the oxide and porosity determination. Figure 2 shows a schematic of the experimental procedure adopted for the preparation of nanocomposite powders and coatings.
3. Results and discussion

3.1. Starting material characterization

Figure 3 shows the SEM micrographs, EDS results and particle size distribution of NiCoCrAlY powder. The NiCoCrAlY powder was gas atomized and has the characteristic spherical shape and quite smooth surface, indicative of good flowability. There are also few satellites, which are typical of the gas atomizing process. The particle size distribution of the powders was in the 5–45 μm range. The chemical composition of the powder obtained by EDS analysis along with the values reported by the manufacturer are shown in table 3.
3.2. Preparation of nanosized CeO\(_2\) powder

The nano-scaled CeO\(_2\) powder has been prepared by high energy ball milling. Nanosized ceria particles have been obtained after 48h milling of as-received ceria powders in the micrometer size range, using ball milling method. Figure 4 shows the 48h ball-milled CeO\(_2\) powder with sub-micron irregular morphology and this type of microstructure was found to be uniform throughout. Small particles can be seen from figure 4(b) and the estimation of the particle size through high magnification FESEM and image analyzer was about 10–30 nm.

The x-ray diffraction patterns of the as-received and milled CeO\(_2\) powders as a function of milling time are shown in figure 5. With increasing the milling time, the initially sharp diffraction peaks of ceria became gradually broader and their intensity was reduced. This effect is mainly ascribed to the decreased crystallite size \[39\]. It should be mentioned that no additional peaks corresponding to any secondary or impurity phase were observed up to 48h of milling indicating high purity of the final sample.

3.3. Preparation of nanocomposite NiCoCrAlY-CeO\(_2\) powders

Figures 6(a)–(f) shows SEM images of NiCoCrAlY-CeO\(_2\) powders at different milling times under condition-A, given in table 2. As can be seen by comparing figure 5 with figure 3, there are differences between the morphology of the powders before and after the milling process. Mechanical milling of the conventional spherical NiCoCrAlY powders led to the formation of irregular and flake-shaped powders. The flakiness of the particles after 48h of milling indicates that the fracturing stage was the controlling mechanism during milling. There are two stages involved in this process, welding and fracturing of the powder particles. It may be relevant to mention here that these powders were retrieved from a still wet, freshly ball-milled mixture. This can be an indication that the added ethanol remained until the end of milling \[40\]. It seems that the presence of PCA (ethanol) reduced the intensity of cold welding during the milling process and almost no welding occurs, as long as the particles are sufficiently covered by PCA. Rana et al obtained similar results \[41, 42\]. They have milled the NiCrAlY powder for 14 h at 300 rpm and using Toluene as PCA. The milled powder was used directly by introducing it into the Low Velocity Oxy Fuel (LVOF) flame to produce nanocrystalline coatings.

Figure 7 shows the particle size distribution for the powders after being milled for different times under condition-A. It can be seen that the size distribution of ball milled powders became broader, as opposite to the narrow range of the conventional NiCoCrAlY powder (figure 3(c)). The variation of powders equivalent diameter with milling time is also given in figure 7. As the milling time increases, it is found that the mean equivalent diameter decreases from about 37 \(\mu\)m in the initial state to 9 \(\mu\)m after the milling process. Due to the mechanical milling, there is a noticeable reduction in particle size and the degree of reduction increases with time. The results are in accordance with SEM observations shown in figure 6.

The mean particle size and powder morphology are important factors that determine the feedstock powder characteristics for thermal spray deposition. Suitable powder conditions for HVOF thermal spraying include spherical or nearly-spherical shape with a size distribution of 10–45 \(\mu\)m \[43\]. Therefore, the milled powders obtained under the milling parameters of A do not have the required characteristics to maintain a continuous

| Element (wt%) | Ni | Co | Cr | Al | Y |
|--------------|----|----|----|----|----|
| EDS results  | Bal.| 23.8 | 18.1 | 11.7 | 0.6 |
| Manufacturer datasheet | Bal.| 20.0–26.0 | 14.0–20.0 | 11.0–14.0 | 0.1–0.8 |

Figure 4. FESEM images of (a) 48h-milled CeO\(_2\) powder and (b) higher magnification of (a).
flow of feedstock powder in the powder feed system. Considering the fact that the morphology and particle size distribution are mainly affected by milling parameters, type B parameters were modified to improve the agglomeration efficiency of the particles. Figures 8(a)–(f) shows SEM images of NiCoCrAlY-CeO₂ powders at different milling times under condition-B.

**Figure 5.** The x-ray diffraction patterns of CeO₂ powder as a function of milling time. The inset is the XRD spectra of the (111) peaks.

**Figure 6.** SEM images of (a) A6, (b) A12, (c) A18, (d) A24, (e) A36 and (f) A48 powders.
It can be clearly seen that the particle morphologies are quite different for different milling parameters. As can be seen, after 3h milling, the starting powder particles underwent a preliminary fracturing and they had a relatively flattened shape (figure 8(a)). Stakes of the laminates are also visible on the flattened particles. Apparently, at this stage of milling, fracturing mechanism dominates over cold welding. Further increasing the milling time led to the agglomeration of small particles into larger ones with rough surfaces (figure 8(b)). The formation of large NiCoCrAlY particles could be related to its FCC crystal structure in which particles are easily deformed and cold welded to each other (under the ball impacts) to form large agglomerates [19]. In practice, the mechanically milled powder particles are deviated from a perfect spherical shape, and in the beginning stages of milling process, a flaky shape may be formed [44]. However, semi-spherical particles were developed after 12h milling time with appropriate morphologies and fine structures for thermal spraying. It seems that the cold welding was the dominant mechanism during the milling period of 6–12 h. Increasing the milling time to 18h, formation of sharp-edged and angular particles was observed (figure 8(f)). Also, it can be seen that the particles became finer which can be attributed to the fracture process activation due to the work-hardening of these particles [45]. The structural evolution and morphological changes observed in the present work after the high energy ball milling treatment is consistent with the results obtained in similar studies. In research by Sharma et al [46], commercially available CoNiCrAlY powder was subjected to ball milling. They have obtained nanocrystalline CoNiCrAlY powders with the particle size of 5–25 μm, after 48 h ball milling. In another study, nanostructured NiCrAlY powders with the average particle size of 126.2 μm were prepared via ball milling [47].

Figure 9 shows the particle size distribution for the powders after being milled for different times under condition-B. It can be seen that the histograms of 3h and 6h-milled powders became broader and the particle size distribution became non-uniform gradually. However, with increasing the milling time, the distribution histograms shift to the left, indicating that the size of the larger particles has been decreased. It seems that by increasing the milling above 12h, as the particles work harden, they fracture more readily.

The effect of particle size and its distribution on the thermally-sprayed coating properties are well understood and it should be carefully controlled to ensure a high-quality coating in terms of low porosity level.
and low oxygen content [48]. The 12h milled powders (sample B12) seem to have suitable morphology and particle size distribution and were selected for HVOF spraying.

Figures 10(a)–(f) shows SEM images of NiCoCrAlY-CeO₂ powders at different milling times under condition-C. In the initial stage of milling (0–12 h), some of the powder particles got trapped between the milling balls however, some of them got escaped through balls. As a result, a part of powder particles received the impact forces from the milling medium and became flattened and the rest maintained their spherical morphology. Further increasing the milling time over 12 h clearly increased the number of flattened particles due to the collision of balls/vial/powder, resulting in the plastic deformation.

Here, unlike the previous cases, it seems that the flattening behavior of the particles was continued till the end of milling time and less fracture was observed in these samples. It is observed that the morphology of powders milled in the presence of stearic acid differed considerably from that of milled with ethanol. It seems that the addition of stearic acid significantly alters the behavior of powder particles during the mechanical milling and delays the stage where fracture process is dominant. This phenomenon can be attributed to the fact that the addition of stearic acid could enhance the lubrication and lead to a decrease in contact area between powder particles [44]. Similar results were reported in other MCrAlY alloys via cryomilling [4, 19].

Figure 11 shows the particle size distribution for the powders after being milled for different times under condition-C. It can be found out that unlike the previous samples, a sharp decrease in particle size was not observed which is due to the different behavior of powder particles during mechanical milling. It is also evident from the plot of equivalent diameter that there is an increase in the mean equivalent diameter which can be attributed to an increase in the particle area (elongated) and reduction of circularity. As shown in SEM images (figure 10), spherical powder particles are deformed and flattened by the ball collisions in which they become elongated. It is apparent that, by increasing the milling time, the distribution histograms slightly shift to the left.
and become broader, compared to the conventional NiCoCrAlY powder. Considering the important powder characteristics, including the morphology and the size distribution, the 24h milled powders (sample C24) were selected for HVOF spraying.

3.4. Distribution of nano-CeO₂

A number of techniques have been developed to incorporate reinforcing agents into the thermally-sprayed coatings. Note that this method should provide a homogeneous distribution of the reinforcing agent to ensure the enhanced properties. Since nanoparticles generally possess high surface energy and they tend to aggregate, achieving a homogeneous nanoparticle dispersoid within the matrix is challenging. Although, the segregation of nanoparticles can be prevented if the milling parameters are correctly adjusted. Figure 12 shows the elemental mapping (Ni, Co, Cr, Al, Y, and Ce) of the selected powders. According to the EDS elemental mapping images, a homogeneous distribution of major alloying elements could be observed in both powders. Also, EDS mapping shows relatively homogeneous distribution of CeO₂ nano-dispersoids in the NiCoCrAlY matrix.

3.5. Structural evaluation of nanocomposite powders

Figure 13 shows the XRD patterns of the as-received NiCoCrAlY and milled powders. The XRD pattern of the as-received NiCoCrAlY shows the presence of two main phases. Four (001), (011), (002), and (111) sharp peaks belong to β phase and the other peaks belong to γ phase. The γ phase is a solid solution of Ni, Co and Cr with a face centered cubic (FCC) crystallographic structure. The β phase is an intermetallic phase with an ordered body centered cubic (BCC) crystallographic structure and (Ni,Co)Al nominal composition [4, 50]. The γ phase (light regions) with β precipitates embedded (dark regions) can also be observed in cross-section SEM image of the as-received powder particle (figure 3(c)). Comparing the XRD patterns of the as-received and milled powders indicate that β-phase precipitates disappear during milling process, probably due to dissolution into the matrix (γ phase). This phenomenon has been reported in the case of several MCrAlY-based alloys [4, 32, 51–53]. The absence of the β phase peaks on the milled powder spectrum could be the result of experiencing severe plastic...
deformation during milling. The microstructural change of the powder particles involves increased energy state of the material and high dislocation density which leads to the instability of crystal structure in the atomic level [52]. Mercier et al [4] reported that the stability of the $\beta$-phase is lower than that of the $\gamma$-phase for the NiCoCrAlY system and the dissolution of $\beta$ phase in the $\gamma$ matrix is more favorable to happen.

Considerable broadening and reduction of intensity of the characteristic Ni, Co, Cr-rich ($\gamma$-phase) peaks can be observed. This could be attributed to the grain refinement, induced micro-stresses, and increased lattice strain during the milling operation [4, 26]. The measurement of the grain size of the milled powder through the peak broadening technique may lead to an inaccuracy which is due to the convolution of the $\gamma$ and the $\beta$ peaks [4, 26]. Hence, in this study, the estimation of the nanoparticle size was conducted on the high magnification FESEM images. High-resolution FESEM in figure 14 has revealed that each nanostructured powder particle comprises of numerous nanoparticles of about 15–25 nm in size.

3.6. Characterization of the coatings

Figure 15 shows the BSE cross-section and SE surface morphology of the as-sprayed conventional and nanocomposite coatings together with their corresponding EDS analysis. The average thickness of the coatings was in the 200–220 μm range. The specimens have similar microstructures, exhibiting the lamellar structure due to the deposition and re-solidification of molten or semi-molten droplets which is the typical feature of the thermally sprayed coatings. However, the cross-sectional image of the conventional coating exhibits a dense structure with lower amount of porosity and oxides compared to other coatings. These microstructural differences can be attributed to the different in-flight temperatures and velocities, powder particles experienced [48]. Compared with other thermal spray processes, the transfer of heat energy (∼2700–3200 K) and kinetic
energy (400–800 m s\(^{-1}\)) is relatively low for HVOF spraying [54]. The observed thin/dark contrast veins in the coatings are known as the oxide stringers. The formation of these oxides may take place during spraying in air (in-flight oxidation) where the metallic particles react with the oxygen present in the flame jet or at the splat surface between spray passes (inter-pass oxidation) [33, 55]. It should be noted that some part of the effect might

Figure 11. Particle size distribution of powders milled for different time as frequency distribution and dependence of equivalent diameter on time.

Figure 12. SEM images of (a) B12, (b) C24 with the respective elemental map.
also be the result of powder particles oxidation during the ball milling process. However, no peaks related to the oxide phases were present in the XRD patterns of the as-milled powders (figure 13). Since the detection limit of the XRD method is considered at about 5%, this might be due to the low volume fraction of these phases and this is true only within the range of detectable of the applied XRD system is this study [56]. Among the three coatings, the coating prepared from the C24 powder has more oxide content than others (figure 15(c)). It seems that the disk-shaped particles with angular morphology favored the chemical reaction between powders (especially the active elements such as Al and Y) and oxygen. Another factor might be related to the nanostructured nature of the powders which would increase the specific heat experienced by a single particle, thereby increasing the in-flight oxidation [57]. When looked at in more detail, it was observed that some elongated particles are present in the coating. Most likely, these particles are originally the as-milled powder particles which did not get fully melted during spraying. On the contrary, the B12 powder shows less surface oxidation due to the semi-spherical morphology which provides good flowability (figure 15(b)). Nevertheless, the coating prepared from the B12 powder has more porosity than others. This could be related to the relatively low deformability of the milled powders, hence promoting the presence of porosity during particle impact [58]. This is also could be the case for the presence of some micro-cracks among the splats of this coating. Presence of the pores and micro-cracks in the coating microstructure could be attributed to insufficient localized plastic deformation during deposition. The total oxide and porosity content of the coatings were estimated by image analysis. The total value was 0.92%, 2.14% and 3.07% for the coatings prepared from the conventional, B12 and C24 powders respectively determined by image analysis.

The images of the top surface of each HVOF coating are illustrated in figures 15(d)–(f). SEM images show that the coatings are rough and continuous without cracks. As can be seen, a higher amount of un-melted powder particles can be seen sitting on the top surface of the conventional coating. If the nanocomposite coatings are compared to the conventional one, a higher amount of melted and re-solidified material can be observed for the nanocomposite coatings. This is true especially for the coating prepared from the C24 powder which this type of powder is more readily melted due to the low thickness of the disk-shaped particles. EDS analysis of the nanocomposite coatings revealed that a high amount of the added nanoceria was present. It is reported in the literature that in the milling process, for ductile/brittle material combinations, the particles of

![Figure 13. XRD patterns for (a) conventional NiCoCrAlY, (b) sample B12 and (c) sample C24 powders with corresponding FWHM values.](image-url)
the brittle powder (CeO$_2$) cut into the metal matrix (NiCoCrAlY), and get surrounded by the metal matrix [36]. Therefore, coatings containing reinforcing particles can be formed by spraying these powders.

Figure 16 displays the fracture-section microstructures of the coatings prepared from the conventional and milled powders. In the case of the conventional coating, un-melted particles are apparent within and on the surface of the coating (as indicated in figure 16(a)). It can be seen that higher amount of un-melted particles are attached to the surface without flattening, compared to the lower deposited layers. This could be attributed to the insufficient particle hammering near the top surface of the coating [59]. Also, the coating contains some pin holes on the surface which can be caused by low impact energy of the un-melted particles. As discussed before, the chemical interactions on the outer surface of the particles which occur during spraying could form an oxide shell, as marked in figure 16(a-2). The fracture-section images of the coating prepared from the B12 powder (figure 16(b)) exhibit porosities and micro-cracks, similar to the cross-section images. It is clear that the porosity content in the coating is dependent on molten state and deformability of the particles and the presence of these defects in this coating could be related to the relatively low deformability of the milled powders. The retained elongated particles are also visible in the fracture-section images of the coating prepared from the C24 powder (figure 16(c-1)). As mentioned before, the disk-shaped particles are more prone to the in-flight oxidation which result in the formation of inter-splat oxide layers in the coating.

4. Conclusions

The focus of this research was to investigate the effect of mechanical milling to prepare a modified NiCoCrAlY powder. Therefore, CeO$_2$ reinforced NiCoCrAlY nanocomposite powders were prepared by mechanical milling process and were examined. The following conclusions can be drawn:
Figure 15. SEM images of cross-section of the coatings prepared from (a) conventional, (b) B12, (c) C24 powders and (d)–(f) their corresponding top surface with EDS analysis performed on (a)–(c).

Figure 16. SEM fracture-sections of the coatings prepared from (a) conventional, (b) B12, and (c) C24 powders alongside with their higher magnification images.
At the first stage of the study, two types of nanocomposite powders were successfully produced by the ball milling process. The evaluation of the milled powders indicated that the morphology and particle size distribution were dependent on the milling conditions.

Mechanical milling under different conditions of the as-received spherical NiCoCrAlY powder led to the formation of semi-sphere (sample B12) and disk-shaped (sample C24) powders. Both powders showed relatively homogeneous distribution of CeO2 nano-dispersoids in the NiCoCrAlY matrix. Considering the important powder characteristics, including the morphology and the size distribution, these powders were selected for HVOF spraying.

After deposition of the as-received and milled powders by the HVOF thermal spray system, the coatings microstructures were evaluated. The specimens had similar microstructures, exhibiting the lamellar structure which is the typical feature of the thermally sprayed coatings. However, some microstructural differences were observed.

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