Chapter 7

Selection of Thermal Spray Process for Bond Coat

7.1. Introduction

Thermal barrier coatings (TBCs) are widely used to enhance thermo-dynamical efficiency in jet engine, gas turbine and internal combustion engines, which are subjected to severe working conditions in an elevated temperature environment [201]. Most TBCs consist of a super alloy substrate, an oxidation-resistant metallic bond coat, and a zirconia-based top coating that is commercially plasma-sprayed onto the bond coat [202]. The plasma sprayed coating presents a strong damage tolerance under the most severe operating conditions, even when a high number of defects are present, such as pores and microcracks. The failure modes related to the material microstructure have been well characterized in the literature [203-204]. The bond coat plays an important role in ensuring structural effectiveness and affording extra adhesion of the top coating to the substrate. Many techniques have been applied to form the bond coat, such as vacuum plasma spray (VPS), low-pressure plasma spray (LPPS), air plasma spray (APS), high-frequency pulse detonation (HEPD), and high velocity oxygen fuel (HVOF) spray [205-209].

Vacuum plasma spraying (VPS) is still the most frequently used thermal spray process for deposition of high temperature oxidation-resistant materials such as MCrAlY alloys onto turbine parts. The process is recognized for providing dense and oxide-poor coatings with superior behavior under thermal and oxidative load [210]. It is, however, also a comparably expensive process since equipment costs are ranging at 2 million US$. Furthermore, the chamber process requires time-consuming evacuation and flooding cycles, which inhibit efficiency [211]. These MCrAlX alloys are also applied to the
superalloy surface by either air plasma-spray (APS), which results in a porous metallic coating with moderate oxidation resistance. It is commonly assumed that the rough surface of the plasma-sprayed bond coat promotes adhesion of the plasma-sprayed ceramic top coat [212]. The typical failure mode of plasma-sprayed TBCs during both engine service and laboratory testing is fracture and delamination of the porous YSZ near to and parallel to the irregular bond coat interface [213-215].

In contrast, high-velocity oxy-fuel (HVOF) spray systems are operated at atmosphere and the investment costs are roughly a tenth compared to VPS and process monitoring is easier. HVOF systems using a Laval nozzle for gas expansion reach gas velocities of over 2000 m/s. The high velocities allow to spray particles in a plastic rather than in a melted state, e.g. for preserving the powder micro structure in the coating. As a result, oxidation of spray material during flight and flattening can be reduced since oxidation can occur only by relatively slow diffusion mechanisms. In spite of the plastic state, the high kinetic energy of the particles still allows flattening by deformation and leads to dense and pore-free coatings. HVOF-sprayed coatings have lower residual tensile stresses than plasma-sprayed coatings and even can have compressive stresses. Till date, the published literatures mainly focused on the failure of TBCs by spallation of the coating at or near a thermally grown oxide (TGO) layer (typically α-alumina), which is created during heat treatment and in service [216]. Limited open literatures are only available related with the effect of thermal spray processes on microstructure and thermo-mechanical properties of the NiCrAlY coatings. Further, acoustic emission technique has been combined with fracture mechanics measurements to investigate the failure behavior of thermally sprayed bond coatings during three point bend testing. Therefore, in this
investigation an attempt has been made to understand the microstructure, mechanical properties such as, hardness, tensile bond strength and fracture behavior of the NiCrAlY coatings combined with acoustic emission and the results are presented in this chapter.

7.2 Thermal Spray Deposition of NiCrAlY Powder

All coatings in this study were fabricated at the Centre for Materials Joining and Research (CEMAJOR), Department of Manufacturing Engineering at the Annamalai University (Annamalai Nagar). The superalloy substrate coupons were grit blasted with Al₂O₃ and then plasma sprayed with NiCrAlY to a nominal thickness of 150 ±10 μm by semi-automatic 40 kW IGBT- based Plasmatron (Model: APSS-II; Make: Ion Arc Technologies; India). For HVOF spraying a commercial manual HVOF (HIPOJET-MEC, Jodpur, India) spraying system was used. The system can be operated with Oxygen, and liquefied petroleum gas (LPG) as fuel gases. For the jet gas shroud, which acts as cooling means and acceleration gas, compressed air can be used. The powder is injected axially into the combustion area and accelerated through a convex- concave nozzle. The parameters used in this investigation for APS and HVOF spraying are shown in Table 7.1.

<table>
<thead>
<tr>
<th>Parameters</th>
<th>APS Units</th>
<th>HVOF Units</th>
</tr>
</thead>
<tbody>
<tr>
<td>Power in kW</td>
<td>25</td>
<td>-</td>
</tr>
<tr>
<td>Standoff distance in mm</td>
<td>110</td>
<td>210</td>
</tr>
<tr>
<td>Primary gas flow rate in lpm (Ar / O₂ )</td>
<td>35</td>
<td>250</td>
</tr>
<tr>
<td>Secondary gas flow rate in lpm</td>
<td>3</td>
<td>--</td>
</tr>
<tr>
<td>LPG Fuel flow rate in lpm</td>
<td>-</td>
<td>80</td>
</tr>
<tr>
<td>Powder flow rate in gpm</td>
<td>30</td>
<td>38</td>
</tr>
<tr>
<td>Carrier gas flow rate in lpm</td>
<td>7</td>
<td>15</td>
</tr>
<tr>
<td>Air flow rate in lpm</td>
<td>-</td>
<td>950</td>
</tr>
</tbody>
</table>
7.3. Coatings Characterization

7.3.1 Effect of thermal spray processes on coatings microstructure

The physical, mechanical and thermo-mechanical properties of thermal spray coatings are governed to a large extent by the nature of the extrinsic microstructural attributes (e.g. pores, cracks, layered morphology). These defect features can supersede the intrinsic material response and can to a certain extent be controlled by process parameters. Thus, for a given material, the choice of the thermal spray process will dictate the microstructure development and, hence, properties. Each process, due to intrinsic differences in the way in which it imparts thermal and momentum transfer to the particles, results in differing signatures in the particle state (namely particle velocity and temperature) and coating build-up process.

Figure 7.1(a) shows the cross-sectional microstructure of plasma spray deposits. APS deposits are very porous and layered with numerous interlamellar pores. The splat thickness is much larger, arising from its much larger droplet. There are “half-penny” shaped pores in the plasma spray deposit; the curved side is on top and the straight or less straight side on bottom. The shape of the pores suggests that the pores were formed due to the expansion of trapped air when the impacting particles were still molten. Splats were also coalesced partially on the interfaces in APS deposits. The coalesced splat/splat boundary is most likely a consequence of metallic bonding and also the under-layer remelting upon the arrival of a subsequent droplet [217].

Fewer and substantially narrower interlamellar pores are observed in the HVOF sprayed coatings. The coating is very dense and the splat/splat boundary is difficult to discern, a typical feature of HVOF coatings are very dense, particles are plastically deformed and densely packed as shown in Fig. 7.1(b).
Fig. 7.1 Optical micrograph of bond coat
Particle/particle boundary grid is clearly seen, no particle coalescence is observed. The formation of interlamellar pores is the consequence of splat curling. The fact that there are very few interlamellar pores in HVOF deposit indicates good adhesion between splats over a large contact area. In the case of plasma spray, the high temperatures and relatively small particle sizes lead to large melting index simultaneously with low Reynolds number. Under these conditions, it is anticipated that flattening ratios will be smaller and potential for splat fragmentation can be higher [218]. Furthermore, it was observed that there is sign of formation of oxide stringers between the splats of plasma spray coating as shown in Fig. 7.1 (c), whereas in HVOF spray coating there is no sign of formation of oxide stringers (Fig. 7.1 (d)).

In the case of HVOF, the large particle size can lead to large splat thickness and lower flattening ratios. This again can contribute to spreading instabilities and resulting fragmentation. In the case of HVOF, the melting index is smaller compared to plasma spray but the smaller particle size distribution and substantially increased flattening ratios result in very thin splats. Under these circumstances, solidification rates for splats are expected to be significantly higher resulting in reduced porosity [219]. Finally, the density of the HVOF is high, it is further evidence that the splat-splat interaction is strong in HVOF. From the porosity analysis results (Table 7.2), it can be inferred that the plasma sprayed coating comparatively possess higher porosity than the HVOF-sprayed coating.

7.3.2 Effect of thermal spray processes on microhardness

The microhardness values of APS and HVOF sprayed bond coats are shown in Fig. 7.2. The relatively porous microstructure of plasma sprayed coatings leads to a lower bond coat hardness values. The APS spray show preponderance of porosity and
interlamellar separations. This is associated with excessive splat fragmentation and poor interlamellar contact due to low velocities. However, HVOF sprayed bond coat shows higher microhardness, because HVOF sprayed bond coat has a dense microstructure [220]. This shows the advantages of the HVOF spray process compared with the APS process. In the HVOF deposit, the average stress was compressive, due to additional high velocity impact that causes plastic deformation in the underlying layers and thus compressive stress (peening effect). This, as a result of high density and better splat-splat adhesion and improved microhardness.

<table>
<thead>
<tr>
<th>Classes</th>
<th>APS Area %</th>
<th>HVOF</th>
<th>APS Pores/sq.mm</th>
<th>HVOF Pores/sq.mm</th>
<th>APS Type</th>
<th>HVOF Type</th>
</tr>
</thead>
<tbody>
<tr>
<td>0 &gt;10 μm</td>
<td>0.78</td>
<td>1.6</td>
<td>821.05</td>
<td>690.68</td>
<td>A 08</td>
<td>A08</td>
</tr>
<tr>
<td>10&gt;25 μm</td>
<td>1.17</td>
<td>0</td>
<td>244.40</td>
<td>0</td>
<td>B 08</td>
<td>-</td>
</tr>
<tr>
<td>25&gt;75 μm</td>
<td>1.46</td>
<td>0</td>
<td>60.82</td>
<td>0</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>75&gt;125 μm</td>
<td>0.83</td>
<td>0</td>
<td>5.63</td>
<td>0</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>&gt;125 μm</td>
<td>1.16</td>
<td>0</td>
<td>1.12</td>
<td>0</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Total</td>
<td>5.4</td>
<td>1.6</td>
<td>1113.47</td>
<td>690.68</td>
<td>Total area measured: 0.0402 sq.mm</td>
<td></td>
</tr>
</tbody>
</table>

Fig. 7.2 Effect of thermal spray processes on coating microhardness of the coating
7.3.3 Effect of spraying processes on surface roughness

There was little difference in the average surface roughness (Ra) of the as-sprayed bond coats (Table 7.3). However, there were marginal differences in the standard deviations, which suggests that there may be differences in the morphology of the surface roughness, if not the magnitude. Characterization of bond coating cross sections by SEM also suggested differences in the bond coat surface topologies. On a qualitative level, the surfaces of APS NiCrAlY appeared to be somewhat more irregular than the surfaces of HVOF NiCrAlY. The deposit microstructures are a direct result of the splat morphologies.

<table>
<thead>
<tr>
<th>Bond coat type</th>
<th>APS</th>
<th>HVOF</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ra (µm)</td>
<td>10.6</td>
<td>9.2</td>
</tr>
<tr>
<td>Standard Deviation</td>
<td>0.3</td>
<td>0.5</td>
</tr>
</tbody>
</table>

Furthermore, the substantially reduced porosity in HVOF can be attributed to limited splat fragmentation and a high degree of flattening. These microstructural attributes are consistent with the interface adhesion results which show that the APS deposits have lowest degree of interface interactions among the two thermal spray processes. The APS offers relatively lower overall impact condition, resulting in the adhesion and interface deficiencies, leads to increased surface roughness values.

7.3.4 Tensile bond strength analysis

The mechanical properties of plasma sprayed coatings, such as adhesion to substrate, can be affected by the processing conditions. It is believed that adhesion is substantially dependent on the microstructure aspects generated under process conditions. Adhesion or bond strength of plasma sprayed coatings is controlled by three main
mechanisms, namely, mechanical anchorage, physical adhesion and chemical adhesion. Among these three mechanisms metallic coatings deposited over the metal substrate bonded together by mechanical locking only. The most important mechanism of plasma sprayed coatings adhesion is mechanical interlocking of the splats to roughness of the substrate.

Surface roughness plays a dominant role in this mechanism. Epitaxy is the second mechanism that can occur if the sprayed material has the same or similar crystal structure [221, 222]. Hence, the bonding strength between the cognate splats (cohesive strength) is higher than the non-cognate splats (adhesion strength). However, physical adhesion and chemical adhesion generally plays only a minor role as an adhesion mechanism in the case of plasma sprayed coatings. Two modes of failure were observed in the test conducted for APS and HVOF spray condition.

From the Fig. 7.3, it is inferred that coatings developed by thermal spray processes, both the coatings exhibits adhesive/cohesive failure due to low bonding strength between the cognate splats. The porosity results suggest that presence of type A and type B pore in the coating matrix due to the processing conditions. Type A pores may be formed as a result of interaction between material particles and the gaseous medium. The type B pores are caused by the splashing of particles on impact with deposited material; or it may be due to voids resulting from the poor deformation of partially melted particles. These pores can have different sizes and exceedingly intricate shapes [72]. Figure 7.4 shows the tensile bond strength results of APS and HVOF sprayed NiCrAlY, higher bond strength was achieved HVOF coating. The reported bond strength values are compared with the powder supplier [221] specified bond strength of the
coating produced by APS process. HVOF coating bond strength values are compared with the commercially available nickel based powder coatings produced by HVOF coating process [222] and comparable results were achieved.

![Figure 7.3](image1.png)

**Fig 7.3** Photograph shows the tensile bond tested specimens

![Figure 7.4](image2.png)

**Fig. 7.4** Comparison of tensile bond strength values

From the perspective of adhesion behavior, it can be asked why the average bond strength of HVOF sprayed bond coats were consistently better than those of BC with APS NiCrAlY, since the only significant difference in the as-deposited coatings was the
bond coat fabrication process. It could be argued that APS NiCrAlY might provide a slight increase in residual stress, since its effective bond strength is likely reduced by the significant amounts of dispersed internal oxides. Hence, it is possible that the APS NiCrAlY experienced a decreased adhesive strength due to oxidation along internal porosity, which may have accelerated BC failure by increasing out-of-plane tensile stresses, whereas as sprayed HVOF BC exhibits compressive residual stress. It could be argued that APS NiCrAlY might provide a marginal increase in residual stress, since its effective bond strength is reduced by the significant amounts of dispersed internal oxides.

Hence, it is possible that the APS NiCrAlY experienced a decreased adhesive strength due to oxidation along internal porosity, which might have accelerated BC failure by increasing out-of-plane tensile stresses, whereas as sprayed HVOF BC exhibits compressive residual stress. The relative lifetimes may have also been affected by subtle differences in bond coat surface morphology or scale adhesion behavior. From the above mentioned points, it can be inferred that the coating sprayed under HVOF spray condition is better because of the denser microstructure as well as of the more integrity of the coating resulted in low porosity and higher bond strength between the splats. Porosity tends to reduce the effective area supporting the load and is detrimental to the strength of the coating [223]. Hence, coatings sprayed by APS process shows reduced adhesive strength.

7.3.5 Evaluation of Young’s modulus of thermally sprayed coatings

To evaluate the Young’s modulus of the sprayed coatings, coated specimens were tested in the three point bend test condition. Because of fact that the tensile testing machine operated at a constant speed, the relationships of load vs. displacement, hits vs.
time and hits vs. displacement could be compared directly. Fig. 7.5 shows Load versus deflection diagram for substrate, APS and HVOF sprayed coating.

![Graph showing Load Vs. Displacement curve for bend tests](image1)

**Fig. 7.5 Load Vs. Displacement curve for bend tests**

![Comparison of Young’s modulus values of coatings](image2)

**Fig. 7.6 Comparison of Young’s modulus value of coatings**

The point at which the macro-cracking in the plasma sprayed coating is indicated by the arrow. This behavior of the coatings can be correlated with their microstructures, since the plasma sprayed coatings can be considered as a collection of weakly bonded tiles, whereas in HVOF spray coatings that have reduced porosity more strongly bonded inter-lamellar boundary would delay the micro crack extension, deflect the cracks before the catastrophic failure. The Young’s modulus values of the substrate, APS and HVOF
coatings were calculated by using the equation nos. 4.1 to 4.3 presented in chapter 4. From the Fig. 7.6, it is understood that HVOF sprayed bond coat exhibits higher Young’s modulus than plasma sprayed bond coat.

### 7.3.6 Acoustic emission analysis

The relationship between the number of hits and displacement of the center of the specimen was found to be an important characteristic. This relationship showed distinct difference between the coatings sprayed by the APS and HVOF spray processes. This is shown in Fig. 7.7 where a plasma sprayed specimen is compared to a HVOF sprayed specimen. From this figure, it can be understood that there exists a great difference in AE production between the plasma sprayed and HVOF sprayed specimens. The HVOF sprayed coatings showed a gradually increasing amount of hits right from the start of the test. This level increased and then became almost constant.

The diagram shows the hits vs. displacement. It has to be noted that the number of hits is not cumulative; each point represents the amount of hits that is produced during a certain time or displacement increment. The coatings sprayed by plasma had a low level of AE after the start of the test. This level was constant or increased only slightly until a certain displacement was reached. At this point the amount of hits abruptly increased. To discover the reason for the abrupt increase of the amount of hits, produced by a plasma sprayed specimen after passing by a certain displacement, a test was carried out where the tensile testing machine was stopped just after the moment that the first high amount of hits appeared. This occurred at a displacement ranging from 1.8 to 2.1 mm, depending spraying process.
Fig. 7.7 Hits vs displacement of thermal spray coatings

Fig. 7.8 Hits Vs. Displacement for 2.1mm

Fig. 7.8 shows Hits vs. Displacement for the same deflection, coatings were bending to same deflection of 2.1 mm. After this, the specimens were prepared for microscopic investigation. The plasma sprayed specimen showed one large crack, running from the top of the coating down to the substrate. Such a large crack not observed in the HVOF sprayed coating but few vertical and interconnected pores were observed. The cracks started at the top of the coating and ran down to the substrate interface.
Fig. 7.9 APS coating shows sudden increase of acoustic burst signals 

Fig. 7.10 HVOF coating shows Linearly increasing acoustic burst signals 

Fig. 7.11 APS coating shows one coarse macrocrack 

Fig. 7.12 HVOF coating shows several cracks 

Figures 7.9-7.12 shows the photographs of the tested coatings, together with their AE diagrams. The distinct difference in AE diagrams was measured, the amount of hits ranged from 100 to 250 for plasma sprayed coatings whereas the number of hits in the HVOF sprayed coatings never exceeded 200. The APS sprayed and HVOF sprayed coatings show distinct differences in the relationship between displacement and amount of hits. This difference may be explained as follows. The HVOF coating shows a very dense network of micro cracks, with an orientation in both vertical and horizontal direction, distributed homogeneously throughout the coatings. The cracks give the HVOF coating certain flexibility. Especially the horizontal micro cracks, which are in fact particle
boundaries, allow the articles to slide over each other over small distances and make it possible for the coatings certain strain before macro crack formation occurs as seen in Fig. 7.12. This in contrary to the plasma sprayed coating, were this micro crack network is almost absent, and macro crack formation in the coating as seen in Fig. 7.11, starts after a certain strain has been exceeded. The amount of AE produced by the HVOF sprayed coatings is caused by the gradual growth of cracks during the loading of the specimens. The amount of hits is therefore also increasing gradually. This in contrary to the plasma sprayed coatings where the macro crack formation, occurring after exceeding a certain strain, caused avalanche of hits.

![Fig. 7.13 Macrograph of crack in APS coating](image1)

![Fig. 7.14 Macrograph of crack in HVOF coating](image2)

The difference in AE production between the plasma sprayed and HVOF sprayed coatings could be explained by the difference in microstructure. Microscopical investigation shows differences in microstructure, the difference probably can be explained by the oxide content of both layers. This oxide content could be explained by oxidation of the coatings after it was sprayed and cooled to room temperature. The oxide content of plasma sprayed coating made the coating more brittle than the HVOF sprayed coating [220]. The cracking oxides caused a high amount of AE. Microscopical investigation showed that the HVOF sprayed
coatings could be strained further than the plasma sprayed coating before cracking occurred. Further, the experiments also showed similar failure characteristics (Figs. 7.13 and 7.14) of the coatings sprayed by thermal spraying method, for a brittle coating/ductile substrate system, vertical cracks firstly appear in coating and propagate towards the substrate interface, and then deflect into interface cracking or delamination with the increase of bending stress.

7.3.7 Fracture Surface Analysis

The elastic modulus (Young's modulus) of materials is one of the most important mechanical properties and it is of prime importance in applied and fundamental fields. It is defined that the elastic modulus is the ratio of stress to strain in the elastic range of a material. It is a measure of the interatomic bonding forces and, thus, the intrinsic characteristic. However, the elastic modulus of thermal sprayed coatings is quite different from comparable bulk materials due to their unique microstructure and inhomogeneity. Furthermore, the elastic modulus of plasma sprayed coatings depends on feedstock particle size (that is, porosity), phase composition, the processing technique, and post-treatment.

When a load is applied to the coating in the direction perpendicular to the coating plane, the force must be transferred from one surface of the coating to the other through the bonded regions between lamellae. Thus there is localized elastic strain at the bonded regions and elastic deflection of the lamellae in the non-bonded regions between them. One problem in determining the elastic modulus of the plasma sprayed coatings is difficulty in obtaining a pure elastic deformation due to splat boundary sliding and/or propagation of cracks and pores. Therefore, the elastic modulus of the plasma sprayed coatings also depends on the method of measurement [47]. Figures. 7.15 and 7.16 are fracture
morphology of coating that may be a typical of plasma-sprayed metallic coatings. It shows distinct lamellar boundaries and columnar grains and coating representing cleavage type catastrophic failure (glass type morphology) due to vertical cracks. A three point bend test accounts for inelastic deformation due to splat boundary sliding and propagation of cracks and pores to evaluate the elastic modulus values, since the plasma sprayed materials can be considered as a collection of weakly bonded tiles. This is also called pseudo-ductility that is due to the cracks opening up.

Fig. 7.15 Fracture surface of APS sprayed bond coat (Low magnification)

Fig. 7.16 Fracture surface of APS sprayed bond coat shows cleavage type failure

Fig. 7.17 Fracture surface of HVOF sprayed bond coat (Low magnification)

Fig. 7.18 Fracture surface of HVOF sprayed bond coat shows ductile mode failure

An important feature to mention on elastic modulus of the plasma sprayed coatings is the anisotropy. That is, the elastic modulus in the direction parallel to the deposition
surface (measured at the cross section of the coating) normally exhibits higher than that in the direction perpendicular to the deposition surface (measured perpendicular to the surface of the coating). The anisotropic elastic behavior and reduced elastic modulus originate from the inter-lamellar pores and intralamellar cracks. The inter-lamellar pores give rise to a significant reduction of the elastic modulus in the in-plane direction, whereas the reduction of the elastic modulus in the cross section direction is attributed to the intra-lamellar cracks.

In addition to the surface area of pores and cracks, the splat boundary area may also contribute to the anisotropy and reduced elastic modulus values in the perpendicular direction [96]. In order to quantify the influence of the microstructure to the anisotropy, it is necessary that the ratio of real contact area between lamellar, the size and the morphology of pores including the microcracks are required to be characterized. HVOF systems using a Laval nozzle for gas expansion reach gas velocities of over 2000 m/s. The high velocities allow to spray particles in a plastic rather than in a melted state, e.g. for preserving the powder microstructure in the coating. As a result, oxidation of spray material during flight and flattening can be reduced since oxidation can occur only by relatively slow diffusion mechanisms. In spite of the plastic state, the high kinetic energy of the particles still allows flattening by deformation and leads to dense and pore-free coatings. HVOF-sprayed coatings have lower residual tensile stresses than plasma-sprayed coatings and even can have compressive stresses. Figures 7.17 and 7.18 shows the fracture surface morphology of HVOF sprayed bond coat exhibits ductile mode of failure.

7.3.8 Thermal expansion behavior

Thermal expansion measurements of bond coatings confirmed that there were significant differences in the thermal expansion behavior of the NiCrAlY processed by
different thermal spraying processes. Figure 7.19 displays the mean coefficient of thermal expansion (CTE) versus temperature of the substrate C263, APS and HVOF bond coats over the temperature interval 100 to 1200 °C.

The plasma sprayed Ni-Cr-Al-Y coating exhibited lower CTE values than HVOF Ni-Cr-Al-Y coating between 100 °C and approximately 1200 °C. However, the Ni-20Cr-10Al-Y experienced a high-temperature phase transformation, as indicated by a 25% increase in mean CTE over the temperature range 900 to 1030 °C. This transformation was reversible and has been previously reported for VPS Ni-22Cr-10Al-0.3Y [42]. The HVOF NiCrAlY did not exhibit an obvious phase transformation of this type. APS NiCrAlY experienced marginal increase in volume due to oxidation along internal porosity, which might have accelerated TBC failure by increasing out-of-plane tensile stresses.

![Graph showing thermal expansion coefficients for C263, APS, and HVOF coatings](image)

Fig. 7.19 Comparison of thermal expansion co-efficient values
The C263 super alloy exhibited the least thermal expansion over the entire range of testing. It is also probable that the thermal expansion behavior of each bond coat is not constant with time at temperature, since the bond coat composition changes significantly during high-temperature exposure. Thus, in order to facilitate more accurate stress models, there is a need to determine the CTE behavior and the time- and temperature-dependent mechanical properties of specific bond coats (on specific substrate alloys) and also been affected by subtle differences in bond coat surface morphology [216-224].

7.4 Summary

(i) Thermal and kinetic energy together controlled the flattening behavior of the particles forming the splats. In the HVOF deposit, the average stress was compressive, due to additional high velocity impact that causes plastic deformation in the underlying layers and thus compressive stress (peening effect), which produces dense microstructure and higher hardness with low surface roughness.

(ii) Approximately, 40% higher bond strength was by the HVOF sprayed coating than APS coating. Tensile bond tested APS coatings shows adhesive / cohesive failure nature of failure, whereas HVOF sprayed tensile bond tested coating failed in an adhesive pattern failure.

(iii) HVOF-sprayed coatings have lower residual tensile stresses than plasma-sprayed coatings and deformed plastically during splat forming, lead to higher strain tolerance. Hence, the fracture surface morphology of HVOF sprayed bond coat exhibits ductile mode of failure.

(iv) Thermal expansion behavior results showed that at higher temperature HVOF sprayed bond coat behaves similar to the substrate material, which will reduce thermal mismatch strain effect and increase TBCs thermal durability.