Fatigue behavior of thermal sprayed WC-CoCr- Steel systems:

Role of process and deposition parameters

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Abstract

Thermal spray deposited WC-CoCr coatings are extensively used for surface protection of wear prone components in a variety of applications. Although the primary purpose of the coating is wear and corrosion protection, many of the coated components are structural systems (aero landing gear, hydraulic cylinders, drive shafts etc.) and as such experience cyclic loading during service and are potentially prone to fatigue failure. It is of interest to ensure that the coating and the application process does not deleteriously affect the fatigue strength of the parent structural metal. It has long been appreciated that the relative fatigue life of a thermal sprayed component can be affected by the residual stresses arising from coating deposition. The magnitude of these stresses can be managed by torch processing parameters and can also be influenced by deposition effects, particularly the deposition temperature. In this study, the effect of both torch operating parameters (particle states) and deposition conditions (notably substrate temperature) were investigated through rotating bending fatigue studies. The results indicate a strong influence of process parameters on relative fatigue life, including credit or debit to the substrate's fatigue life measured via rotating bend beam studies. Damage progression within the substrate was further explored by stripping the coating off part way through fatigue testing, revealing a delay in the onset of substrate damage with more fatigue resistant coatings but no benefit with coatings with

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inadequate properties. The results indicate that compressive residual stress and adequate load bearing capability of the coating (both controlled by torch and deposition parameters) delays onset of substrate damage, enabling fatigue credit of the coated component.

1. Introduction

Thermal Spray (TS) is a widely utilized coating technology that services a variety of industries including aerospace, heavy machinery, and power generation, wherein a feedstock material (often in powder form) is melted and propelled towards a surface, producing a shower of particles that impact and rapidly solidify on the surface, (known as splats), which form a coating through successive layers. Within the wide menu of TS processes, High Velocity Oxy-Fuel (HVOF) deposition has received particular attention for its ability to produce very dense (>99% [1, 2]) and well bonded coatings. HVOF processing also has the capability to induce compressive residual stresses within the deposit due to solid-state peening achieved by the high kinetic energy particle impact during deposition, reducing or neutralizing the tensile stresses generated by splat quenching[3, 4]. As such, HVOF processing is often employed for the deposition of dense hard facing materials, such as WC-CoCr, to protect wear prone components of various functions. Characterization and optimization of these coatings' wear and corrosion behavior have been extensively studied as a function of processing and resultant coating properties [1, 5, 6]. However, in the case of a coated component subject to cyclic loading, such as aero landing gear or hydraulic cylinder rods, understanding the fatigue behavior of the coating and substrate as an integrated system is critical for coating design.

Numerous academic fatigue studies of TS coated specimens have been carried out, however there appears to be no consensus on whether an enhancement [7-10] or debit [7, 11-14] in fatigue life is generated with coating application. Industry data is rarely available for such studies. Since TS coating properties are highly dependent on the processing parameters used [5, 15], an assessment with one coating process parameter does not offer a complete picture on the fatigue of TS coated specimens [13, 16]. With parametric studies of coating processing, it has been

reported that the residual stress state of the coating has a large influence on the relative fatigue life of coated specimens [12, 16, 17], as well as coating properties such as coating toughness [10]. Residual stress of thermal spray coatings have also been shown to have considerable influence in contact fatigue scenarios [18]. The presence of residual stresses throughout a material prone to fatigue loading has been considered a critical factor for the fatigue life of uncoated components [19, 20], as the superposition of the applied stress and residual stress throughout a fatigued body leads to local stress intensifications. Such stress intensifications are then prone to fatigue crack initiation and growth as described by Paris' Law[21]. Other investigations into the effect of substrate preparation prior to coating deposition, such as shot peening or grit blasting, have also been studied as residual stress contributors in attempts to mitigate fatigue debits incurred by coating application [11] or to isolate the coating's influence on fatigue life [7].

Manipulation of TS coating residual stress via processing has been the subject of many previous studies, where torch operating parameters, feedstock characteristics, and deposition rates have all been demonstrated as having large roles in determining the magnitude of coating deposition stresses [3-5, 22]. In addition to these deposition stresses, thermal stress arising from differences in the coating and substrate's coefficient of thermal expansion (CTE) as the specimen cools back to ambient temperature contributes to the coating residual stress. As a deposition parameter, variations in cooling and part geometry and size can cause significant differences in substrate temperature with the otherwise same processing parameters, producing considerable thermal stress as the coated part cools back to ambient temperature. Summated with the deposition stress to produce the final residual stress, the thermal stress component of a coating is a key

contributor, particularly when large thermal expansion mismatches between the substrate and coating exist and/or high substrate temperatures are experienced during spraying.

This paper seeks to address several unanswered questions on the fatigue behavior of HVOF WC-CoCr coated steel through studies of spray conditions and depositions conditions. In this paper, “torch parameter influence” will refer to differences in the coating process achievable by varying parameters such as the torch’s fuel or oxygen flow and the subsequent results of thermal and kinetic energy transfer onto the feedstock material, independent of how the particles interact with a substrate. “Deposition condition influence” will refer to how the particles are directed to and interact substrate material, particularly the substrate temperature. Some salient questions that will be addressed through these studies include:

- Isolate the role of spray process conditions and deposition temperatures on the fatigue response of the coated system.
- Impact of substrate surface preparation on fatigue response and the potential modification to the effect upon coating application.
- Understanding fatigue damage accumulation within a coated specimen and the role of process conditions on damage progression.

2. Experimental Methods

Coatings for the investigation of torch parameter influence on fatigue life were produced using a DJ-2600 HVOF torch (Oerlikon-Metco, Westbury, NY) with commercially available JK 120H WC-CoCr powder (Kennametal-Stellite, Goshen IN). The “DJ-A” and “DJ-B” conditions were selected from an experimental design where in-situ curvature-based stress measurements of

coatings produced with different levels of process gas flow, oxygen to fuel ratio, stand-off distance, and powder feed rate were performed [23]. The DJ-A and DJ-B conditions were selected for use based on the large differences in formation and residual stresses between the coatings sprayed with these conditions. Coatings for the investigation of deposition condition influence on fatigue life were produced with a JP-5220 HVOF torch system (Praxair Surface Technologies, Indianapolis, IN) with commercially available WOKA 3652 WC-CoCr powder (Oerlikon-Metco, Westbury, NY). Torch parameters for these conditions are shown in Table 1

For planar samples, a torch rastering speed of 1000m/s was used with a step size between strokes of 3mm and 5mm for the DJ and JP, respectively.

For cylindrical fatigue samples, the torch rastering speed was 30mm/s and 50mm/s for the DJ and JP, respectively, in line with the center axis as the samples were rotated at 630 RPM in order to keep the same pitch between coating passes as the planar samples and avoid barber-poling. Shielding was used on the fatigue samples in order to eliminate over-spray. Cooling air was directed at planar and cylindrical samples in different manners but kept consistent for the specified geometry.

Planar geometry samples consisted of 25.4 x 225 mm steel beams that were mounted to an *in-situ* curvature measurement device in order to calculate deposition, thermal, and residual stresses, with back-side substrate temperature monitored by contact thermocouples. Such a device and calculations were first developed for thermal spray by Kuroda and Clyne [24], with modelling further expanded upon by Tsui and Clyne [25], and compared to other residual stress measurement methods by Sampath and Matejicek [26, 27]. Detailed descriptions of device use

and calculations can be found in references [3, 5, 28]. Micrographs were prepared from these beams and examined in back scatter mode under SEM (Hitachi TM3000)

Rotating Bend Fatigue (RBF) specimens were machined out of cold drawn 1018 steel to the nominal dimensions shown in Figure 1. Rotating bend testing was chosen as the fatigue method since a reversible, higher stress amplitude would be directed to the surface and coating, amplifying the coating parameter influence on specimen fatigue life. 1018 steel was selected since it has similar modulus and thermal expansion coefficient to higher strength steels but a lower fatigue strength that allows the coating parameter influence on fatigue life to be further amplified. All samples were heat treated within a vacuum furnace at 900°C for a dwell time of 2hrs and allowed to cool within the furnace to ambient temperature in order to remove work hardening and reduce the cold drawn grain structure within the steel to produce a stress free steel with a more isotropic grain distribution.

The planar steel beams and selected RBF samples were grit blasted using an alumina grit. Samples were cleaned in an acetone bath with ultrasonic agitation for cleaning prior to coating deposition to ensure no trapped grit or dust would be included in the coating-substrate interface.

Surface temperature of the necks of the RBF samples during spraying were monitored by an IR thermal camera (FLIR Systems, Wilsonville, OR), with manipulation of the temperature being accomplished by robotically varying the time between coating strokes on the samples. Cooling air was directed at the neck of the samples during spraying and cooling. Overspray of the samples was eliminated by securing fitted caps machined to size onto the ends of the substrates doing the spray process. Coatings were sprayed to a nominal thickness of 165µm, for a total

diameter increase of 330 μ m. No grinding or polishing of the coating after deposition was done in order to not alter the roughness of the coating after deposition.

WC-CoCr coating stripping was performed using a solution of Potassium Sodium Tartrate (Rochelle Salt) and Sodium Carbonate in distilled water at concentrations of 50 g/l and 150 g/l, respectively. A D.C. voltage of 3-4 volts was applied to the sample and an immersed steel electrode, producing a current of 1-1.3 Amps. The solution was continuously stirred and kept ~60°C during the stripping process, with periodic light brushing of the coated area to expedite the stripping process. After coating removal, the RBF samples were washed with distilled water and dried prior to fatigue testing. Similar procedures have been used to effectively strip WC-CoCr coatings [29, 30]

Fatigue testing was performed using a RBF-200 model rotating beam fatigue testing machine (Fatigue Dynamics Inc., Dearborn MI, now System Integrators LLC, Glendale AZ). Samples were tested at a frequency of ~100Hz at varying applied moments until failure or severe deformation triggered the motor's cutoff switch while the cycle counter recorded the number of cycles to failure. It should be noted that in order to assume a stress amplitude value for the surface of the RBF specimens, the increase in specimen diameter due to coating thickness is treated as if the coating is identical to the substrate (i.e., coating modulus = 205 GPa), which is not an ideal assumption [8-10]. Micrographs of fatigue fracture surfaces were taken using either an optical stereoscope or SEM in backscatter mode.

Vickers hardness of the coatings sprayed on planar substrates was measured on mounted and polished cross sections at a load of 300g for 15s. Indentation modulus measurements were taken on the top surfaces of polished coatings using an instrumented indenter with a Berkovich tip (Micro-Materials Limited, Wrexham Technology Park, Wrexham, UK). Elastic recovery of the

coating upon unloading was used to calculate the indentation modulus, as described by Oliver and Pharr [31].

~~It should be noted that in order to assume a stress amplitude value for the surface of the RBF specimens, the increase in specimen diameter due to coating thickness is treated as if the coating is identical to the substrate, which is not an ideal assumption [8-10]. However, in the case of most coated component design, the coating is not treated as a load carrier but its thickness is taken into the measurement of the part's geometry. Thus all stress amplitudes on the fatigue specimens are reported as if composed of the substrate material alone but include both the substrate and coating thickness.~~

3. Results

3.1 Quantification of Residual Stresses

In order to quantify the differences in process induced residual stress that can be achieved by torch and deposition conditions, beam curvature measurement was used to monitor the deposition of WC-CoCr. Figure 2(a) shows the raw curvature-time data of conditions DJ-A and DJ-B sprayed onto low carbon steel planar beams, with deposition and cooling regimes noted, [to a thickness of 230µm and 200µm, respectively](#). A clear difference in the curvature evolution between the two samples are seen, with positive curvature accumulated during deposition with coating DJ-A and a slightly negative curvature accumulated for DJ-B. This indicates the DJ-A coating having a tensile deposition stress, as calculated by the Stoney formula [32] inset in Figure 2(a). In contrast, the DJ-B coating has a compressive deposition stress which indicates that the peening effect of high kinetic energy particles has overcome the tensile quenching stresses occurring during coating deposition. Average backside substrate temperature during

deposition was measured to be approximately 190°C for DJ-A and 260°C for DJ-B, contributing to a slightly higher thermal stress contribution in DJ-B. Final residual stress in the coating is a combined effect of process and thermal stresses shown in Figure 2(b).

In the second set of experiments a single spray torch parameter was used but two sets of coatings were processed to impart different thermal stresses via changing substrate temperatures. This was achieved by changing the pause time, and hence cooling time, between coating deposition layers using robotic programming, resulting in two coating types; nominally low temperature (LT) and high temperature (HT). Both the LT and HT coatings accumulated similar positive curvature (and hence tensile deposition stress), though at different rates with respect to time, shown in Figure 3(a), with a thickness of 215µm and 225µm, respectively. The average back side substrate temperature stabilized at approximately 350°C for the High Temperature (HT) coating and 200°C for the Low Temperature (LT). As such, a larger compressive thermal stress is produced in the HT coating by the higher deposition temperatures. Both the LT and HT coating had a similar deposition stress value, making the residual stress primarily different by the thermal stress component. Calculated values for the deposition, thermal, and combined residual stress calculated from the beam curvature measurements for the LT and HT coatings are shown in Figure 3(b), using the Stoney formula.

3.2 Coating Microstructure and Properties

Micrographs of the four coatings are shown in Figure 4. The torch parameter difference between DJ-A and DJ-B make a noticeable difference in the size and frequency of coating porosity, with

DJ-B showing a denser microstructure than DJ-A. The LT and HT coatings show a similar porosity character relative to each other.

Hardness measurements of the DJ-A and DJ-B coatings revealed a large difference between the two coatings, with measured values of 1170 and 1330 Kgf/mm², respectively. It is clear to see that the difference in torch operating parameters and thus in-flight particle state had a significant influence on coating hardness for DJ-A and DJ-B, in agreement with literature findings [5, 23] . Detailed property and performance measurements of DJ-A and DJ-B have been previously published as coatings D1 and D9, respectively, in ref [23]. To assess the mechanical property influence from deposition parameters, hardness and indentation moduli of the HT and LT WC-CoCr coatings were measured and are shown in Figure 5. Here it can be seen that only a slight difference in coating properties can be observed with the HT coating having the greater hardness and indentation modulus, though the difference is within the standard deviation of the measurements. This points to the time between coating deposition passes having a much smaller effect on the HT and LT coating mechanical properties (if any) as compared to the influence particle state can have, as is the case for the DJ-A and DJ-B coatings

3.3 Torch parameter influence on fatigue life

The DJ-A and DJ-B conditions were applied to the steel RBF specimens and were fatigue tested until failure. In order to avoid the complexities in fatigue behavior arising from surface compressive residual stress, roughness, and grain distortion that can be induced by grit blasting, the coatings were applied without prior grit blasting of the substrate. No coating delamination was observed during fatigue testing until final failure. Substrate temperature was measured during deposition by IR camera to reach 215°C for the DJ-A coating and 325°C for the DJ-B coating application. The stress vs. number of cycles to failure curve (S-N) for steel (S), DJ-A coated, and DJ-B coated samples are shown in Figure 6. It can be seen that the DJ-B coating adds a significant enhancement to the fatigue life of the coated specimen, while the DJ-A coating has a slight debit to the fatigue life compared to the uncoated specimens.

3.4 Thermal stress influence on fatigue life

As with the previous example, a set of RBF samples were coated without grit blasting using the LT and HT conditions under a single torch parameter condition. As noted earlier, these two conditions impose different levels of net residual stress in the coating without a significant change to the microstructure, Vickers hardness, or indentation modulus. No coating delamination was observed while spraying onto the substrates or during fatigue testing, which confirmed adequate bonding. A set of specimens where the steel was grit blasted prior to coating deposition was also tested with the same coatings. The S-N for the bare substrate, LT coated, and HT coated samples, without and with grit blasting prior to spraying, are shown in Figure 7 (a) and (b), respectively. It is clear to see that there is a large difference between the coated samples' fatigue life depending on which coating is applied to the steel and if grit blasting was used.

In the cases of substrates that were not grit blasted, the HT coated samples (S + HT) produced a large increase in fatigue life vs. the uncoated steel. The LT coated samples yielded a fatigue debit relative to the uncoated steel, indicating a high sensitivity of the coating application and surface deposition temperature on expected specimen fatigue life.

From Figure 7(b), it is evident that grit blasting slightly increased the fatigue life relative to the annealed steel samples, which can be attributed to the compressive stresses and grain distortion induced into the surface of the sample from the blasting process. The samples with the LT coating applied to grit blasted substrates did not show significant fatigue life difference from the grit blasted only samples, indicating that the grit blasting was able to improve the fatigue life of the LT coating and cancelled the negative effect of the LT coating in the non-grit blasted samples. ~~Although compared to the LT coated samples without grit blasting (S + LT), the difference between a fatigue debit or credit with LT coating deposition relative to the annealed steel samples is shown to be dependent on grit blasting prior to coating application.~~ The samples with grit blasting and the HT coating showed an increase in fatigue life relative to the grit blasted only specimens, with a higher fatigue life than the HT coated samples that were not grit blasted, further improving fatigue life. Thus it can be said that grit blasting enhanced the fatigue life of the coated and non-coated specimens in all cases relative to the non-grit blasted counterparts.

3.5 Fractographic observation of fatigue failures

A typical fracture surface of the non-grit blasted and coated samples is shown in Figure 8(a), noted within is the fatigue crack initiation and final fracture site. Coating delamination was

observed on many specimens adjacent to the final fracture site, though no adhesive coating failure was observed adjacent to fatigue crack initiation sites, confirming an adequate coating bond strength during fatigue testing, even without grit blasting. Figure 8(b) shows a more detailed image of the crack initiation site, where cohesive coating failure was frequently found in the adjacent area.

Fracture surfaces of the sample grit blasted prior to LT WC-CoCr coating deposition are shown in Figure 9(a). The coating shows a severe amount of in-plane cracks, as well as complete cohesive delamination adjacent to the fatigue crack initiation site in the substrate. The fracture surface of a coating on a non-grit blasted substrate is shown in Figure 9(b), with an inset highlighting a crack that passed through the coating and substrate. Cohesive cracking within the coating is again visible, yet this instance shows a crack path passing through the substrate and into the coating.

3.6 Partial fatiguing

In order to assess how the coating and substrate deteriorate during fatigue testing, coated samples from the LT and HT set were fatigue tested for ~~certain~~ portions of the expected fatigue life and then halted from testing prior to fracture. The WC-CoCr coating was then removed by the stripping processes described in the experimental methods and resumed fatigue testing until failure as a bare substrate. No grit blasting of the substrates was done prior to coating application in these cases. [With the LT and HT coated samples requiring very different stress amplitudes to break within the high cycle fatigue regime, a direct comparison of pre-fatiguing was not possible. Instead the stress amplitude of the stripped specimens was held at~250 MPa for comparison of damage from the pre-fatiguing.](#) The results of this testing is shown in Figure 10 (a) for a sample with the LT coating and Figure 10 (b) for a sample with the HT coating, both with the relevant

data from Figure 7 shown in grey. Sample a, with the LT WC-CoCr coating applied, was fatigued to approximately 50% of the expected linearly measured fatigue life at that stress. , ~~which correlates to 95% of its expected logarithmically measured fatigue life.~~ After pausing fatigue testing and stripping the coating off, sample a' was fatigued at the same load until failure, falling short of the expected fatigue life for the uncoated steel samples, indicating significant damage to the substrate had occurred during prior fatigue testing while the coating was present. Sample b with the HT WC-CoCr coating applied was fatigue tested to the similar portion of expected fatigue life as sample a (50% linear, ~~95% logarithmic~~), though at a higher load and longer time due to the HT coated sample's prolonged life. After pausing and stripping the coating off, sample b' was fatigue tested to failure, reaching a life equivalent to that of the baseline annealed 1018 steel material, suggesting that no significant damage accumulated within the substrate during the prior fatigue testing.

In order to confirm that the spray process itself did not alter the substrate's fatigue life, specimens were sprayed with the HT and LT coating, stripped, and fatigue tested until failure. No difference in fatigue life was observed between the bare steel substrates (S) and samples that underwent coating, stripping, and fatigue testing.

4. Discussion

The results presented above points to the importance of process induced residual stress states of coatings on the fatigue response of a WC-CoCr coated steel system. In one instance, the stress states were modulated through changing particle kinetic and thermal energies via process conditions (DJ-A and DJ-B coatings). In this instance, the rotating bend fatigue data suggests that coatings processed with compressive deposition stress through high particle kinetic energy resulted in fatigue credit of the system while coatings sprayed under tensile evolving stresses

showed a slight decrease in specimen fatigue life. ~~no change to the substrate fatigue response.~~
The higher peening stress associated with DJ-B also aided in the densification and hardness of the coating, which can also be contributing factors to an enhanced fatigue life. By changing deposition temperatures for constant spray conditions, different levels of thermal stresses can be imposed onto the coating-substrate system. This was demonstrated through the analysis of the LT and HT coatings. Here the coating densities, hardness, and indentation modulus were similar but the final stress states of the coated samples were significantly different.

Though direct comparison of coating residual stress as measured by beam curvature cannot be directly translated to cylindrical geometry, calculation of deposition, thermal, and residual stress in coatings on both flat and cylindrical geometries have been established by Tsui and Clyne [25, 33]. Figure 11 shows the magnitudes of the expected longitudinal thermal stress component to the residual stress for the LT and HT coatings, calculated using the equations described in [33] and a deposition temperature of 50°C and 150°C for the LT and HT coatings, respectively. A coating modulus of 1.5 times that of steel (~300GPa) is used to demonstrate an estimate of the WC-CoCr coated steel system, with lower coating moduli (1 and 0.5 times that of steel) also shown to demonstrate how a coating with lower stiffness would influence the thermal stress magnitude. Depending on the modulus of the coating after deposition, the magnitude of the stress imposed in both the coating and substrate can vary, with a stiffer coating producing higher stresses. This demonstrates the high sensitivity of the coating's stress state upon the thermal stress, with a significantly lower stress induced in the substrate as compared to the coating for this scenario. It is important to note that a coating material with a different CTE and mechanical properties would result in different expected coating-substrate stress profiles, such as other common wear resistant thermal spray coatings like Cr₃C₂-NiCr which has a more closely

matching CTE to steel and therefore a less compressive thermal stress, potentially leading to differences in fatigue life. Calculation of an expected deposition stress contribution using the Tsui-Clyne model for the RBF specimens cannot be assumed to be the same as the planar specimens due to differences in relative surface speed, expected angular particle impact, and different substrate surface temperatures experienced by depositing splats. However, calculations indicate approximately a 10 to 1 ratio for the magnitude of deposition stresses experienced by the coating and substrate, respectively, within the dimensions of these RBF samples. This again underlines the critical importance of processing on coating residual stress.

The resulting stress profile within the substrate and coating upon application of a bending moment to the RBF sample is a superposition of both the residual stress and the applied stresses, determining the local stress intensity that can initiate and drive fatigue cracks. The magnitude of the applied stress throughout the specimen will depend on the coating's thickness and modulus, altering the experienced stresses within the substrate as compared to an uncoated sample under the same load. Specimen curvature calculated using statics equations of pure bending of a coated cylinder (an assumption for the thinnest diameter of the RBF specimen) take into account the moduli difference and dimensions of the coating and substrate under a linear elastic assumption as such

$$K = \frac{64 M}{E_s \pi D_s^4 + E_c \pi (D_{s+c}^4 - D_s^4)} \quad (1)$$

Where K is the curvature of the specimen, M is the applied moment, E is the modulus of the coating or substrate, and D is the diameter of the substrate or substrate with coating. Calculating

the curvature of the specimen then allows for calculation of the local stresses throughout the specimen

$$\sigma_s = E_s K R \quad (2)$$

$$\sigma_c = E_c K R \quad (3)$$

Where R is the radius throughout the specimen. Figure 12 shows the calculated stress profile under a single moment using these equations and the same relative values for the coating modulus as in Figure 11. It is clear to see that coating addition lowers the local-mean stress amplitude within the substrate by the coating accommodating a portion of the load, depending on the modulus of the coating. However, there may also be some stress concentrations within the substrate due to non-homogeneity (cracks, poor adhesion) within the coating. It is also important to note that the load carrying capability of the coating will likely be lower under a tensile stress compared to a compressive stress due to the large amount of microstructural defects within the coatings (pores, micro-cracks, inter-lamellar boundaries etc.). This will also play a role in crack initiation and propagation, where tensile fatigue stresses typically drive crack growth. Finally, it can also be stated that as the stiffness of a coating decreases through damage accumulation, such as thinning from wear, corrosion, or fatigue cracking, more load is distributed back to the substrate and thus raises both the mean and local stress within the substrate[34].

The expected residual and applied stress profiles and coating properties can now be utilized to explain the fatigue performance of these coated specimens. As evident for samples b' in Figure 10(b) testing a significant portion of the expected HT coated sample's fatigue life did not accumulate significant damage within the substrate, indicating that the stiffness of the coating

adequately reduced the stress amplitude experienced by the substrate, delaying damage accumulation. The higher hardness and more compressive residual stress in the coating, such as the case for DJ-B and HT, contributes to the integrity of the coating being maintained for a longer period of fatigue testing and delays substrate damage. The opposite case was observed with the LT coating in samples a and a', with the higher amount of coating tensile stress not being able to maintain the coating integrity, with the possibility of a lower coating stiffness not adequately alleviating the stress experienced by the substrate as well. It would be expected that the DJ-B and DJ-A coatings would behave in a similar manner, with the DJ-B coating carrying more load through its denser microstructure and experiencing a slower rate of damage accumulation from its more compressive residual stress state and higher hardness as compared to the more porous, more tensile, and less hard DJ-A coating. Thus it is the load carrying ability and damage accumulation rate of a coating that determines fatigue life enhancement or debit. Additional residual stresses due to grit blasting or shot peening would also superimpose to the stress profile, offering a benefit to the substrate and thus to the coated specimen's fatigue life as seen in Figure 7. These stresses produced via grit blasting are largely left unaltered by the coating process heat, which is evident by the samples that were coated and stripped prior to fatigue testing. It is also suggested from the stripped coating samples without any grit blasting that the high kinetic energy particles that initially impact the substrate surface do not impart a significant fatigue life benefit like grit blasting or shot peening. Thus it can be said that the state of the substrate is largely left unaltered by the coating processes in this scenario. This may not be the case for more thermally sensitive or softer substrates, such as tempered aluminum or copper for example.

Experimental studies, as well as modelling [35], have reported on the behavior of crack advancement through two well bonded materials of the same modulus, with crack advancement in the softer material occurring when the crack is initiated in the harder material, and crack blunting when an initiated crack within the softer material approaches the interface of a harder material. In the case of a WC-CoCr coating perfectly bonded to a 1018 steel substrate, it is possible that crack advancement from the coating into the substrate can occur. However, the weaker TS coating-substrate interface, which relies on mechanical interlocking and the anisotropic nature of thermal spray coatings inherent in their deposition method [36] offers different crack pathways than the through-thickness direction, producing the in plane and interfacial branching directions the coating crack takes in Figure 9(b). The large amount of in-plane (cohesive) coating cracks observed on the fracture surfaces of the majority of the RBF samples agree with this preferred crack pathway, with primary fatigue crack initiation occurring in the substrate. It can be hypothesized that the coating can act to delay fatigue damage initiation and propagation within the substrate if adequate stress reduction is supplied to the substrate. The hardness and residual stress of the coating determines its ability to resist damage due to cyclic loading. However, the presence and growth of the cracks within the coating and their orientation does not offer a pathway for crack propagation directly into the substrate. Instead, the failure of the specimens is due fatigue crack initiation and growth within the substrate. ~~yet does not increase the likely hood for premature deleterious coating to substrate crack advancement in these scenarios and is instead driven by substrate fatigue crack initiation and growth.~~

5. Conclusion

Manipulation of residual stresses and coating properties in TS WC-CoCr deposited by HVOF was achieved and quantified on planar steel samples through torch operating parameters and

deposition temperature. *In-situ* beam curvature measurements were used to monitor both the stress evolution during layer by layer coating deposition and during the post-deposition cooling sequence. ~~Measured properties of the coatings produced by different torch parameters indicated difference in coating hardness, while properties on both planar and cylindrical samples with different substrate temperatures showed only slight differences in modulus and hardness.~~ Steel rotating bend fatigue specimens were tested with and without coatings applied following the torch and substrate condition variation as described above. It was found that coatings with higher compressive residual stress, hardness, and microstructural density enhanced fatigue life of the coated specimens. Coatings with tensile residual stress and somewhat weaker properties (hardness/stiffness) incurred a fatigue debit in the system.

Coated specimens that were fatigue tested for a portion of their expected fatigue life, ~~stripped of their coating,~~ and retested ~~as bare substrates following stripping of the coating~~ provided insights as to the coating quality on the ~~specimen's~~ fatigue behavior. Samples with tensile coating stresses and lower hardness weakened the system resulting in fatigue deficit of the substrate while coatings with strong compressive residual stress and better process induced deposit integrity showed fatigue credit to the underlying steel. Fracture surface analysis revealed fatigue crack initiation within substrates, with the adjacent coating containing considerable cohesive damage. ~~Grit blasting was revealed to marginally improve the fatigue life of the substrates and eliminate the fatigue debit in substrates with the LT coating applied, while further increasing the fatigue life of the HT coated samples.~~ Analytical calculations of the longitudinal thermal stress differences between ~~coatings~~ the LT and HT coatings assuming cylindrical geometry were made using the Tsui-Clyne model, revealing the considerable influence of expected differences in the residual stress profile within the coating and substrate. Statics calculations for pure bending of a

coated cylinder were also made in order to demonstrate the expected load bearing capacity of the coating and subsequent stress reduction within the substrate. Finally, a proposed mechanism for failure in these samples expects that the residual stress and coating integrity resulting from processing are key factors in determining the fatigue response of the system. ~~If the coating is able to support a portion of the load, invariably the system will show improvement in fatigue behavior, where as if the coating integrity is such that it cannot support any load in the composite, the system will show fatigue debit.~~ This study thus provides a frame work for process design for optimal performance of thermal spray coated components subjected to fatigue loading.

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References

- [1] A. Ghabchi, T. Varis, E. Turunen, T. Suhonen, X. Liu, S.-P. Hannula, Behavior of HVOF WC-10Co4Cr coatings with different carbide size in fine and coarse particle abrasion, *Journal of thermal spray technology*, 19 (2010) 368-377.
- [2] J. Kawakita, S. Kuroda, T. Kodama, Evaluation of through-porosity of HVOF sprayed coating, *Surface and Coatings Technology*, 166 (2003) 17-23.
- [3] A. Valarezo, S. Sampath, An integrated assessment of process-microstructure-property relationships for thermal-sprayed NiCr coatings, *Journal of thermal spray technology*, 20 (2011) 1244-1258.
- [4] S. Kuroda, Y. Tashiro, H. Yumoto, S. Taira, H. Fukanuma, S. Tobe, Peening action and residual stresses in high-velocity oxygen fuel thermal spraying of 316L stainless steel, *Journal of thermal spray technology*, 10 (2001) 367-374.
- [5] T. Varis, T. Suhonen, A. Ghabchi, A. Valarezo, S. Sampath, X. Liu, S.-P. Hannula, Formation Mechanisms, Structure, and Properties of HVOF-Sprayed WC-CoCr Coatings: An Approach Toward Process Maps, *Journal of Thermal Spray Technology*, 23 (2014) 1009-1018.
- [6] D. Toma, W. Brandl, G. Marginean, Wear and corrosion behaviour of thermally sprayed cermet coatings, *Surface and Coatings Technology*, 138 (2001) 149-158.
- [7] O. Kovářik, J. Siegl, Z. Procházka, Fatigue behavior of bodies with thermally sprayed metallic and ceramic deposits, *Journal of thermal spray technology*, 17 (2008) 525-532.
- [8] A. Ibrahim, C. Berndt, The effect of high-velocity oxygen fuel, thermally sprayed WC-Co coatings on the high-cycle fatigue of aluminium alloy and steel, *Journal of materials science*, 33 (1998) 3095-3100.
- [9] A. Ibrahim, C. Berndt, Fatigue and deformation of HVOF sprayed WC-Co coatings and hard chrome plating, *Materials Science and Engineering: A*, 456 (2007) 114-119.
- [10] A. Ibrahim, R. Lima, C. Berndt, B. Marple, Fatigue and mechanical properties of nanostructured and conventional titania (TiO₂) thermal spray coatings, *Surface and Coatings Technology*, 201 (2007) 7589-7596.
- [11] H. Voorwald, R. Souza, W. Pigatin, M. Cioffi, Evaluation of WC-17Co and WC-10Co-4Cr thermal spray coatings by HVOF on the fatigue and corrosion strength of AISI 4340 steel, *Surface and Coatings Technology*, 190 (2005) 155-164.
- [12] R. McGrann, D. Greving, J. Shadley, E. Rybicki, T. Kruecke, B. Bodger, The effect of coating residual stress on the fatigue life of thermal spray-coated steel and aluminum, *Surface and Coatings Technology*, 108 (1998) 59-64.
- [13] M. Costa, M. Venditti, H.J.C. Voorwald, M. Cioffi, T. Cruz, Effect of WC-10% Co-4% Cr coating on the Ti-6Al-4V alloy fatigue strength, *Materials Science and Engineering: A*, 507 (2009) 29-36.
- [14] Y. Itoh, M. Saitoh, K. Takaki, K. Fujiyama, Effect of high-temperature protective coatings on fatigue lives of nickel-based superalloys, *Fatigue & Fracture of Engineering Materials & Structures*, 24 (2001) 843-854.
- [15] A. Valarezo, W.B. Choi, W. Chi, A. Gouldstone, S. Sampath, Process control and characterization of NiCr coatings by HVOF-DJ2700 system: a process map approach, *Journal of thermal spray technology*, 19 (2010) 852-865.
- [16] J.A.M. de Camargo, H.J. Cornelis, V.M.O.H. Cioffi, M.Y.P. Costa, Coating residual stress effects on fatigue performance of 7050-T7451 aluminum alloy, *Surface and coatings technology*, 201 (2007) 9448-9455.
- [17] A. Vackel, G. Dwivedi, S. Sampath, Structurally Integrated, Damage-Tolerant, Thermal Spray Coatings, *JOM*, DOI 67 (2015) 1540-1553 ~~1-14~~.
- [18] R. Ahmed, M. Hadfield, Mechanisms of fatigue failure in thermal spray coatings, *Journal of thermal spray technology*, 11 (2002) 333-349.

- [19] S. Kikuchi, Y. Nakahara, J. Komotori, Fatigue properties of gas nitrided austenitic stainless steel pre-treated with fine particle peening, *International Journal of Fatigue*, 32 (2010) 403-410.
- [20] S. Hassani-Gangaraj, A. Moridi, M. Guagliano, A. Ghidini, M. Boniardi, The effect of nitriding, severe shot peening and their combination on the fatigue behavior and micro-structure of a low-alloy steel, *International Journal of Fatigue*, 62 (2014) 67-76.
- [21] P. Paris, F. Erdogan, A critical analysis of crack propagation laws, *Journal of Fluids Engineering*, 85 (1963) 528-533.
- [22] K. Shinoda, J. Colmenares-Angulo, A. Valarezo, S. Sampath, Effect of deposition rate on the stress evolution of plasma-sprayed yttria-stabilized zirconia, *Journal of thermal spray technology*, 21 (2012) 1224-1233.
- [23] A. Vackel, G. Dwivedi, S. Sampath, Influence of in-flight particle states on the particle peening intensity and related properties of HVOF WC-CoCr coatings, *Thermal Spray Bulletin*, 8 (2015) 32-41.
- [24] S. Kuroda, T. Clyne, The quenching stress in thermally sprayed coatings, *Thin solid films*, 200 (1991) 49-66.
- [25] Y. Tsui, T. Clyne, An analytical model for predicting residual stresses in progressively deposited coatings Part 1: Planar geometry, *Thin Solid Films*, 306 (1997) 23-33.
- [26] J. Matejcek, S. Sampath, In situ measurement of residual stresses and elastic moduli in thermal sprayed coatings: Part 1: apparatus and analysis, *Acta Materialia*, 51 (2003) 863-872.
- [27] O. Kesler, J. Matejcek, S. Sampath, S. Suresh, T. Gnaeupel-Herold, P. Brand, H. Prask, Measurement of residual stress in plasma-sprayed metallic, ceramic and composite coatings, *Materials Science and Engineering: A*, 257 (1998) 215-224.
- [28] T. Suhonen, T. Varis, S. Dosta, M. Torrell, J. Guilemany, Residual stress development in cold sprayed Al, Cu and Ti coatings, *Acta Materialia*, 61 (2013) 6329-6337.
- [29] B.D. Sartwell, K.O. Legg, J. Schell, J. Sauer, P. Natishan, Validation of HVOF WC/Co thermal spray coatings as a replacement for hard chrome plating on aircraft landing gear, DTIC Document, 2004.
- [30] R. Menini, N.B. Salah, R. Nciri, Stripping methods studies for HVOF WC-10Co-4Cr coating removal, *Journal of materials engineering and performance*, 13 (2004) 185-194.
- [31] W.C. Oliver, G.M. Pharr, An improved technique for determining hardness and elastic modulus using load and displacement sensing indentation experiments, *Journal of materials research*, 7 (1992) 1564-1583.
- [32] G.G. Stoney, The tension of metallic films deposited by electrolysis, *Proceedings of the Royal Society of London. Series A, Containing Papers of a Mathematical and Physical Character*, 82 (1909) 172-175.
- [33] Y. Tsui, T. Clyne, An analytical model for predicting residual stresses in progressively deposited coatings Part 2: Cylindrical geometry, *Thin Solid Films*, 306 (1997) 34-51.
- [34] A. Vackel, T. Nakamura, S. Sampath, Mechanical Behavior of Spray-Coated Metallic Laminates, *Journal of Thermal Spray Technology*, DOI 25 (2016) 1009-1019.
- [35] Y. Sugimura, P. Lim, C. Shih, S. Suresh, Fracture normal to a bimaterial interface: effects of plasticity on crack-tip shielding and amplification, *Acta Metallurgica et Materialia*, 43 (1995) 1157-1169.
- [36] S. Usmani, S. Sampath, D.L. Houck, D. Lee, Effect of carbide grain size on the sliding and abrasive wear behavior of thermally sprayed WC-Co coatings, *Tribology transactions*, 40 (1997) 470-478.

Table 1: Torch parameters of the conditions used for depositing the DJ-A, DJ-B, and LT/HT coatings along with measured average particle temperature and velocity using accuraspray.

Condition	Kerosene Flow [l/hr]	Hydrogen Flow [slpm]	Air Flow [slpm]	Oxygen Flow [slpm]	Stand Off Distance [mm]	Powder	Feed Rate [g/min]	Particle Temperature [°C]	Particle Velocity [m/s]
DJ-A	-	636	334	188	254	JK 120H	30	1849	609
DJ-B	-	731	408	276	229	JK 120H	30	1918	731
LT, HT	24	-	-	940	330	WOKA 3652	65	1827	811

Figure 1: Rotating bend fatigue specimen nominal dimensions, with portion coated by WC-CoCr indicated.

Figure 2: (a) Raw curvature-time data for the DJ-A and DJ-B conditions sprayed onto low carbon steel beams. Undulations during deposition represent the torch rastering over the beam, with positive curvature accumulation for DJ-A and negative curvature accumulation for DJ-B. In-set shows the Stoney formula and appropriate variables for stress calculations. (b) Calculated deposition, thermal, and residual stress values from beam curvature of the WC-CoCr coatings. Note that the deposition and thermal stress for DJ-A were nearly the same, making a near-zero residual stress in the coating.

Figure 3 (a) Smoothed curvature-time data for the High Temperature (HT) and Low Temperature (LT) coating conditions sprayed onto low carbon steel beams. Note that similar net curvature change occurs for the deposition of the two coatings, though they occur over different time scales due to the robotically controlled pause time between coating deposition passes. A large difference in curvature change between the two coatings is due to the different substrate temperatures during spraying; a direct result of the pause time between deposition passes as well. (b) Calculated Deposition, Thermal, and Residual stresses of the HT and LT coatings. A similar deposition stress but largely different thermal stress summate to a large residual stress difference in the coatings, with HT being in compression and LT being nearly neutral

Figure 4: Back scatter SEM images of the coating microstructures cut from the steel beams used to measure stress. Comparison of the DJ samples shows larger and more frequent porosity in DJ-A than in DJ-B. LT and HT show similar porosity character, differing only in substrate temperature

Figure 5: Indentation modulus and cross section hardness of the HT and LT WC-CoCr coatings on planar beams

Figure 6: Stress amplitude vs. cycles to failure for 1018 steel (S) with DJ-A and DJ-B WC-CoCr coatings applied (Referred to as S+DJA and S+DJB, respectively, in order to avoid confusion). The DJ-B coating offers significant improvement to the specimen fatigue life, whereas the DJ-A coating produces a slight deficit. Arrows indicate runout.

Figure 7: (a) Stress amplitude vs. cycles to failure for 1018 steel (S) with Low and High Temperature WC-CoCr coated steel (S+LT and S+HT, respectively). (b) Stress amplitude vs. cycles to failure of RF specimens that were grit blasted (S + GB) with and without LT and HT coating deposition. Arrows indicate runout. The HT coating shows fatigue improvement in all cases, whereas the LT coating either offers a deficit or minimal change from the substrate behavior.

Figure 8: (a) ~~Optical image~~ **Light micrograph** of typical fatigue fracture surface with fatigue crack initiation, final fracture site, and coating de-bonding from final fracture indicate. SEM of the highlighted region of initiation site shown in (b) with cohesive coating failure and fatigue crack initiation site. Both images are of a single S + HT sample

Figure 9: SEM images of fatigue fracture surface of coated samples (a) grit blasted prior to coating deposition, with fatigue crack initiation site and cohesive coating cracks indicated and (b) non-grit blasted prior to coating deposition with similar cohesive coating cracks, as well as a substrate crack. Inset image highlights the observed crack passing through the coating-substrate interface, though not a result of crack driving into the substrate from the coating. Images are those sprayed with the LT condition

Figure 10: Stress amplitude vs. cycles to failure for coated samples that were partially fatigue tested, stripped, and tested as uncoated until failure for a sample with (a) LT coating and (b) HT coating. Related samples from Figure 3 are included in grey as reference

Figure 11: Calculated longitudinal thermal stresses of a cylinder using the Tsui-Clyne model [33] for the Low Temperature (LT) and High Temperature (HT) coatings with coating modulus at 0.5, 1.0, and 1.5 times that of the steel substrate. Deposition temperatures of 50°C and 150°C were used for the LT and HT coatings, respectively, and ambient temperature of 25°C

Substrate: $E = 205 \text{ GPa}$, $\nu = 0.29$, Radius = 3.12 mm, $\alpha = 12 \text{ } \mu\text{m/m } ^\circ\text{C}$

Coating: $\nu = 0.23$, Thickness = 0.165 mm, $\alpha = 6 \text{ } \mu\text{m/m } ^\circ\text{C}$

Figure 12: Calculated applied stress profile from pure bending moment (6.78 Nm, ~~60 in-lb~~) of a cylinder with coating modulus at 0.5, 1.0, and 1.5 times that of the steel substrate. Dimensions used were the same as in Figure 11

1

2a

2b

3a

3b

7a

7b

8a

8b

9a

9b

10a

10b

12