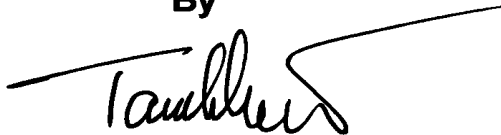# A LOOK AT HARD CHROME REPLACEMENT

REPORT # ANM-112N-01-02   JULY 2001

REVISION A   JANUARY 2002

By

Terry Khaled, Ph.D.
Chief S / T Advisor, Metallurgy
Federal Aviation Administration
3960 Paramount Boulevard.
Lakewood, CA 90712
(562) 627-5267
terry.khaled@faa.gov

Abstract

For decades, hard chrome (Cr) plating has been utilized in applications that require corrosion protection and/or wear resistance and also in the salvage industry, for rebuilding worn or undersized components. While Cr plating offers certain benefits, it also has some decided disadvantages. In recent years, these disadvantages have led to a global interest in seeking a replacement. Tungsten carbide (WC)- cobalt (Co) and WC-Co-Cr coatings, thermally sprayed by the HVOF (high velocity oxy-fuel) process, have emerged as the likely replacement candidates. There are several significant efforts underway to evaluate HVOF coatings and compare them to Cr plating. Some of the data obtained in the various efforts are inconsistent, even conflicting. In this paper, the results of the efforts, for which information is available, are reviewed in some detail. Potential culprits in the inconsistent and conflicting data issue are explored. It is argued that performance of any deposit, be that Cr plating or some HVOF coating, is a sensitive function of processing parameters and not a generic material property of that deposit. Suggestions are offered to improve confidence and minimize conflict in the data collected in ongoing and future efforts. It is felt that some collaborative research work is required.

1.0 INTRODUCTION

1.1 Issue

For decades, hard chrome (Cr) plating has been utilized in applications that require corrosion protection and/or wear resistance and also in the salvage industry, for rebuilding worn or undersized components. While Cr plating offers certain benefits, it also has some decided disadvantages. In recent years, these disadvantages have led to a global interest in seeking a replacement. Tungsten carbide (WC)- cobalt (Co) and WC-Co-Cr coatings, thermally sprayed by the HVOF (high velocity oxy-fuel) process, have emerged as the likely replacement candidates. There are several significant efforts underway to evaluate HVOF coatings and compare them to Cr plating.

1.2 Purpose

The purpose of this document is fourfold. First, to present a technical review of the hard Cr replacement efforts for which information is available. Second, to assess and attempt to reconcile some of the results that appear to be conflicting. Third, to enhance awareness of certain aspects of plating and coating performance. Fourth, to suggest means to improve confidence and minimize conflict in the data collected in ongoing and future efforts.

1.3 Limitations

The views presented herein are those of the author and, as such, they should not be construed as reflecting a unified FAA position. FAA applicants are hereby advised that the responsibility for setting requirements and for approval belongs to the cognizant FAA certification organizations.

2.0 BACKGROUND

2.1 Engineering (Hard) Cr Plating

2.1.1 General

Hard Cr is plated from the same solutions used for decorative Cr plating (chromic acid solutions that contain sulfates, chromates, fluorides and/or carbonates); longer times are required for the hard Cr. Plating is performed using insoluble Pb-Sb (or Pb-Sn) anodes at temperatures around 120-140 F. Conforming anodes may be used to enable plating in holes and in other restricted areas. Plating thickness for hard Cr is in the 0.001 to 0.020 in. range (typically, 0.008 to 0.010 in.), but thicker deposits are possible; decorative Cr seldom exceeds 0.00005 in. The plating has high hardness, typically in excess of HV 900 (HRC 67) and also provides low friction against most metals and alloys. It can be applied directly to the substrate and, when required, ground to size and/or polished to a high surface finish. In what follows, hard Cr plating will be referred to as Cr plating.

2.1.2 Defects in Cr Deposits

Cr, deposited under normal conditions, is ordinarily cracked; the cracks are referred to as microcracks. With each layer deposited, old cracks are partially bridged and additional cracks develop. The cracks in consecutive layers, however, tend not to line up with each other, resulting in a random pattern. Several mils need to be deposited, before all cracks are covered to an extent that there is no path from top to bottom of the deposit; in general, 1.5 mils (0.0015 in.) of deposit is sufficient. Microcracking is best explained by the hydride formation theory. The theory assumes that the initial deposit contains hydrides of Cr (CrH, CrH_2 and Cr_2H). These hydrides can have a face centered cubic or hexagonal structure. The latter, however, is the one favored under normal plating conditions. The Cr hydrides thus formed are unstable and they eventually break down to Cr (body centered cubic) and hydrogen (H). This break-up is accompanied by up to 15% volume shrinkage and the formation of microcracks that are akin to mud cracking. Shrinkage stresses, however, are not completely relaxed by microcracking. As a result, the deposit and a substrate layer immediately underneath that deposit are left under residual tensile stresses. For deposits with sufficient thickness (say, 0.0015 in. or more), microcracks usually do not expose the substrate and, as such, they are not viewed as potentially damaging. It has even been theorized that microcracks are beneficial in that they offer reservoirs for lubricant retention, thereby reducing run-off and providing a constant long-term supply of that lubricant. By contrast, the residual tensile stresses lead to fatigue life debit and can also give rise to adverse effects in sustained-load performance.

There is another type of cracks in Cr deposits; namely, macrocracks. Macrocracks, while less prevalent than microcracks, are longer and wider; typically a macrocrack is at least 0.0015 in. in length. Macrocracks extend from the top of the deposit to the interface and, as such, they can adversely affect corrosion resistance.

Macrocracks result from rapid part heating or cooling, pre-coating residual stresses in the substrate, contaminated plating solutions and/or rectifier ripple. Other defects can also exist in hard Cr deposits; examples are poor adhesion/flaking, porosity and nodular deposits. These can develop from improper surface preparation and/or if the plating parameters are not controlled and maintained within optimum ranges.

2.1.3 The Trouble with Cr

In recent years, a global interest has developed to seek a replacement for Cr plating. The main reason for this is the hazardous nature of the plating solutions. These solutions contain hexavalent Cr and, as such, they present a health risk and require special precautions to use and to dispose of. These precautions are placing an ever-increasing toll on the use of Cr plating.

Another reason for seeking a Cr replacement is the lengthy processing times required. Cr plating is a slow process; deposition rates of about 0.001 in. per hour are typical. Thus, during the deposition of thick coatings, the plating parameters must be monitored and maintained within optimum ranges for lengthy periods of time; otherwise, plating defects will develop. This is a rather difficult task. Additional processing time needs to be budgeted, to accomplish two other treatments, associated with electroplating high strength steels; namely, stress relief and the post-plate baking. These treatments are mandated by concern over sustained load cracking. Sustained load cracking is a phenomenon that is adversely affected by the presence of high levels of tensile stresses, residual or applied, and also by high hydrogen concentrations. Stress relief is required if the components have been subjected to machining, grinding and similar operations, before plating. It is intended to reduce the level of any residual tensile stresses that might have been generated by these operations, thereby mitigating hydrogen effects. According to QQ-C-320, a min. of 3 hr @ 375 F is required for stress relief. Apart from mitigating hydrogen effects, a reduction in residual tensile stresses would tend to improve fatigue performance; as such, stress relief may also be beneficial for coatings that do not involve the generation of hydrogen, such as thermal spray. Stress relief aside, post-plate baking is required for hydrogen embrittlement relief. The intent of this baking is to reduce, through diffusion, the high surface concentration of hydrogen, picked up during plating, thereby improving resistance to sustained load cracking. According to QQ-C-320, a min. of 12 hr @ 375 F is required for this relief and a 200 hr sustained load coupon test is required to verify compliance. Lengthy processing times, apart from adding to cost, can be objectionable in the salvage industry, due to turn-around considerations.

A third reason for seeking a replacement is the fatigue debit associated with Cr plating. There is ongoing research to remedy the situation. For example, it is reported that a reduction in the fatigue debit, together with improvements in corrosion and wear performance, could be realized by utilizing an electroless Ni underplate. [1] While future improvements are possible, the hydride theory, presented earlier, suggests that the deposition of Cr will always involve a fatigue debit. It is to be noted, however, that most plated and thermally sprayed coatings involve a fatigue debt.

A fourth reason that is occasionally cited in the literature is the pre-plating surface preparation required to achieve adequate adhesion. Such surface preparation, however, is as important for other plated and thermally sprayed coatings as it is for Cr plating. Regardless of the method by which a deposit is applied, be that plating or thermal spray, improper surface preparation would result in less-than-adequate adhesion which, in turn, can lead to flaking and/or under-plate corrosion.

A fifth reason that is frequently cited in the literature is the inconsistent corrosion protection afforded by Cr plating. This inconsistent protection, however, is not an inherent property of Cr plating. Rather, it is brought about by the presence of macrocracks, flaking and similar defects that expose the substrate. The occurrence and severity of such defects are controlled by processing details; e.g., pre-cleaning, current density and bath chemistry and temperature. These details, no doubt, vary from one plating house to the other. In their never-ending quest for the low bidder, users are amenable to changing plating houses. Frequently, due to processing differences, the incidence of defects in the "new" plating is not the same as in the "old" one, giving rise to differences in corrosion protection. The problem is further compounded by processing changes within the same house, in the never-ending quest to cut cost. It is thought that a similar situation exists for thermally sprayed coatings.

2.2 Thermally Sprayed Coatings

2.2.1 General

Thermal spray is a generic term for a group of processes in which coating materials are heated to or near their melting points. The resulting droplets, consisting of molten or partially molten material, are accelerated in a gas stream and projected against the surface to be coated (the substrate). On impact, the droplets flow into thin lamellae (AKA splats), adhering to the surface, overlapping and interlocking as they solidify. The total coating thickness is usually generated in multiple passes. The coating material may be metallic, inorganic (ceramic or carbide), polymeric, or composite, in the

form of powder, wire or rod. Thermal spray processes belong to the broader field of vacuum and controlled atmosphere coatings, which also includes the CVD, PVD (and its variants) and diffusion coating processes.

2.2.2 Classification of Thermal Spray Processes

Thermal spray processes can be classified in two groups. The first group includes processes that employ an electrical heat source; namely, the plasma and wire-arc processes. The second includes processes that employ a chemical (combustion) heat source; namely, the flame, detonation gun and super detonation gun (D-Gun and SDG, respectively) and the HVOF processes. In what follows, the focus will be on combustion processes.

2.2.3 The Combustion Processes

2.2.3.1 Historical: Flame spraying was introduced in the 1910's and, as such, it was the very first thermal spray process introduced. In the 1950's, Union Carbide (now Praxair Surface Technologies) developed and commercialized the D-Gun and later the SDG; initially for the application of WC coatings and later for other types of coatings. HVOF (Jet Kote) was introduced in the early 1980's. Initially, HVOF was developed as a competitor to the D-Gun and SDG processes for the application of WC coatings. Later on, the process was further developed to enable the application of metallic coatings as well.

2.2.3.2 Operation: To be sure, all combustion processes utilize an oxidant and a fuel and, as such, they could all be labeled as oxy-fuel processes. In the flame process, combustion of the oxy-fuel mixture takes place in ambient air, without much confinement. In the HVOF process, by contrast, combustion takes place within the confines of a combustion chamber. The combustion products then accelerate to high velocities upon exiting through a straight nozzle. Even higher velocities can be attained if the combustion products are passed through a converging-diverging (Laval) nozzle. In both the flame and HVOF processes, the combustion products heat and propel the coating material towards the substrate. The D-Gun and SDG operate on a somewhat different principle. The powder is mixed with the oxy-fuel mixture in a chamber and then ignited (detonated) by a spark plug; detonation makes the process noisy. The resulting sound wave heats and propels the coating material toward the substrate. The chamber is then purged with nitrogen and the cycle is repeated; more cycles per second (7 or so) are used in the SDG, compared to the D-Gun (4 or so). D-Gun, SDG and HVOF are strictly powder processes; reportedly, however, some wire HVOF equipment exist. By contrast, the flame process can be used with powder, wire or rod.

2.2.3.3 Powders for HVOF, D-Gun and SDG: In as much as WC coatings are involved, three materials have been used extensively; namely, WC-12%Co, WC-18%Co and WC-10%Co-4%Cr. Sometimes, the Co content referenced in various publications is lower than the nominal values indicated here; e.g., WC-17%Co, instead of WC-18%Co. These lower than nominal values reflect the fact that the Co content in the deposited coating, due to evaporation, is less than its counterpart in the initial powder. WC is very hard, typically in excess of HV 2000. Lower overall hardness values are to be expected when WC is mixed with Co (or Co and Cr); the higher the Co (or Co+Cr) percent, the lower the overall hardness. WC-Co coatings with hardness values in excess of HV 1000 can be obtained.

2.2.3.4 Process Comparisons: The inherent advantage of HVOF, D-Gun and SDG over all other spray processes is the ability to spray semi-molten droplets at high velocities. The result is higher bond strength, reduced oxide content and lower porosity.[2] Thicker coatings can also be applied by these methods.[3] Due to its continuous combustion nature, however, HVOF can lead to undesirable substrate heating and property degradation, if the processing parameters are not properly controlled. With proper process control and powder selection, HVOF, D-Gun and SDG can induce compressive residual stresses; all the other thermal spray processes, with the exception of some plasma techniques, tend to induce residual tensile stresses. It is worth noting that D-Gun and SDG coatings can be applied only at third party contract facilities, where the loud operation can be accommodated in proper enclosures. Potential thermal spray users tend to view this restriction as a serious limitation. By contrast, HVOF equipment are marketed by several companies and are available at an ever-increasing number of facilities around the globe. Furthermore, HVOF powders, covering a wide range of metallic, carbide, ceramic and composite materials, are also available from a number of producers. These features make HVOF attractive to potential users.

2.3 Adhesion & Bond Strength (Appendix A)

Plating and thermal spray specifications generally require some form of quantitative and/or qualitative adhesion verification. In the case of plated metallic deposits, the existing specifications only require some form of qualitative testing for adhesion verification. Thermal spray coating specifications, on the other hand, frequently require both qualitative and quantitative testing for this verification. Bending is the most widely specified form of qualitative testing, for both plated and thermally sprayed deposits. Quantitative adhesion verification, for thermally sprayed coatings, is exclusively through the tension test of ASTM C 633 or its derivatives. This test, however, is limited by the strength of the adhesive used. The HVOF coatings of interest here, when properly applied, can develop bond strengths greater than the

tensile capabilities of the commercially available adhesives. In such cases, the otherwise quantitative test is rendered qualitative.

2.4 Cr Plating Replacement Efforts (Appendix B)

Significant efforts are now underway to evaluate the merits of HVOF WC-Co and WC-Co-Cr coatings in comparison to Cr plating. These include efforts by the Department of Defense (DOD), Boeing-Long Beach Division (LBD), Boeing-Seattle (Boeing) and Airbus Industries. There are two separate DOD efforts. The first, and more involved, is that of the HCAT (Hard Chrome Alternatives Team). The second is that of the NAWC (Naval Air Warfare Center), and it is more oriented towards certain needs of naval aviation. A summary of the more important results of the various efforts is presented below.

2.4.1 Corrosion Testing

DOD-HCAT and LBD corrosion testing per ASTM B117 indicated virtually no difference between HVOF WC-17% Co coating and Cr plating. ASTM B117 testing by Airbus Industries revealed that HVOF WC-10% Co-4% Cr coating outperformed the HVOF WC-17% Co coating which, in turn, outperformed Cr plating. In atmospheric corrosion testing, DOD-HCAT results indicate that WC-17% Co coating performed markedly better than Cr plating. Thus, it seems reasonable to conclude that, in as much as corrosion protection is involved, performance of HVOF WC-10% Co-4% Cr and WC-17% Co coatings is at least as good as, if not better than, that of Cr plating.

2.4.2 Stress Corrosion Testing

DOD-NAWC testing revealed that Cr plating is superior to the HVOF WC-17% Co coating. This is because the latter tended to spall and expose the substrate, as a result of the tensile loading associated with stress corrosion testing.

2.4.3 Corrosion Fatigue Testing

DOD-NAWC testing indicated that the HVOF WC-17% Co coating is superior to Cr plating which, in turn, is superior to the bare steel.

2.4.4 Hardness Testing

DOD-HCAT microhardness testing showed the HVOF WC-17% Co coating to be somewhat harder than Cr plating.

2.4.5 Wear Testing

DOD-HCAT abrasive wear testing revealed that the HVOF WC-17% Co coating is superior to Cr plating, suggesting that abrasive wear is influenced by hardness. LBD block on ring wear testing revealed that the HVOF WC-17% Co coating is superior to Cr plating. This result, however, is suspect, due to poor adhesion of Cr plating on some of the specimens. Airbus Wear testing revealed that HVOF WC-17% Co coatings are at least as good as Cr plating; the nature of the wear test performed, however, was not disclosed.

2.4.6 Bond Strength Testing

LBD and Boeing performed bond strength testing based on ASTM C633. The LBD results were deemed inaccurate, as a result of improper test set-up. The Boeing results are not available. As pointed earlier, bond strength testing based on ASTM C633 has its limitations. These limitations render the test incapable of yielding quantitative information about the bond strengths of the coatings of interest here.

2.4.7 Tensile Testing

LBD Tensile testing indicated that the heat input associated with the improper application of HVOF coatings can degrade the strength properties of the substrate. There is no mention of tensile testing in any of the other Cr replacement efforts.

2.4.8 Fatigue Testing

Fatigue testing in all efforts consisted of obtaining S-N curves for coated and bare steels. The coatings compared were Cr plating and thermally sprayed WC-17% Co. In addition to WC-17% Co, Boeing also tested WC-10% Co-4% Cr. The main thermal spray method investigated was HVOF. In addition to HVOF, LBD and Boeing investigated D-Gun and SDG. The steels used in the various efforts were different. AISI 4340 was used in the DOD-HCAT and Boeing efforts. AerMet 100 was used in the DOD-NAWC effort. 300M was used in the LBD effort. E35NCD16, a French ultrahigh strength steel,

was used in the Airbus Industries effort. Focusing on Cr plating and HVOF coatings, highlights of the test results are presented below.

(a) The results of the DOD-HCAT and NAWC efforts indicate that Cr plating leads to a substantial fatigue life debit, compared to uncoated steel. The HVOF coating involved a much less pronounced debit. In either case, the debit appeared to increase at lower stress levels.

(b) Airbus Industries test results, using shot peened specimens, indicate that both Cr plating and the HVOF coating involve a fatigue debit. The debits associated with both coatings, however, are more-or-less the same. Actually, the curves depicted indicate that Cr involves a somewhat lesser debit than the HVOF coating.

(c) LBD test results indicate that both Cr plating and the HVOF coating involve a fatigue debit. For both coatings, the fatigue debit generally increased at lower stresses. Increasing the thickness of the deposit was accompanied by an increase in the Cr plating debit and a decrease in that of the HVOF coating. The data further show that for thinner coatings, especially in the low cycle (high stress) regime, Cr plating can involve somewhat lesser debits than those of HVOF coatings. The trend is reversed, however, for thicker coatings and/or in the high cycle (low stress) regime.

(d) The Boeing test results seem to indicate that HVOF coatings can give rise either to a fatigue life debit or credit. The data further suggest that, specimens HVOF coated by different processors can have different fatigue lives, in spite of the fact that the powders were procured to the same specification. Finally, the data indicate that Cr plating is always associated with a life debit.

3.0 DISCUSSION

The reasons for seeking an alternative to Cr plating, cited in section 2.1.3, are no doubt valid. It appears, however, that the general belief is that thermal spray coatings are the answer to all Cr plating concerns. Nothing could be farther from reality. Only time will show that thermal spray coatings have their own inconsistencies and difficulties. Some problem areas are presented in section 3.1. Section 3.2 discusses the importance of adhesion and bond strength, as determining parameters in coating performance. Finally, section 3.3 explores some potential culprits in the conflicting fatigue data, obtained in the various efforts. Throughout this section and elsewhere, the term "coating performance" is used. This is an abbreviation for "coated substrate performance."

3.1 Limitations of the HVOF Process

3.1.1 A Line of Site Process: An inherent limitation of all thermal spray processes, including HVOF, is their line-of-sight nature; that is to say, only those surfaces that the gun or torch could "see" may be coated. In addition, spray gun size limitations can restrict HVOF applicability to certain components. As a result, not all of the currently Cr plated components, or areas thereof, would be amenable to coating by thermal spray processes. Examples include designs that require the coating of holes, recesses, keyways and the like. For such applications, either the Cr plating has to be retained or some alternative to thermal spray has to be sought. One such alternative is electroless nickel plating. [4] Many in the aviation industry, however, have a mind set against electroless Ni. The reasons for such a mindset include concerns about toxicity of the plating solutions, [5] the difficulty of controlling the plating process and inconsistent corrosion protection. One can not help but notice that similar concerns triggered the quest to seek a replacement to Cr plating in the first place. This seems to suggest that the mindset is really against all plating processes.

3.1.2 Substrate Heating: The fast deposition rates and continuous combustion associated with HVOF can lead to undesirable substrate heating and property degradation, if the processing parameters are not properly controlled. The LBD tensile data show that this, indeed, can take place. Thus, supplier qualification and lot acceptance should include tests to guard against that possibility. In this regard, tensile testing can be a powerful quality assurance tool to detect gross substrate heating. Microhardness testing and metallographic examination would be equally powerful tools in detecting local heating effects just underneath the coating.

3.1.3 Porosity: All thermal spray coatings, to varying degrees, have inherent porosity and, as such, they should be sealed for maximum corrosion resistance. Sealants are either organic (polymeric) or inorganic. Organic sealants include oils, waxes, epoxies, and phenolics. Epoxy and other organic primers may also be used as sealants. Inorganic sealants are mainly silicates. As is the case with castings, sealant application may be by brush, dipping, spray or by vacuum impregnation. The sealant must be compatible with the environment in which the component will operate. Porosity and the presence of sealants can interfere with bond strength testing based on ASTM C633. More on this in section 3.2.

3.1.4 Property Variations: Performance of any deposit, in a given application, is not inherent to the material. Rather, it depends on the processing details as well as on the particulars of deposited material. For example, performance of a

particular WC-Co coating material deposited by HVOF is not likely to be equivalent to that of the same material deposited by, say, the SDG or plasma process. This would be the case even if the same powder and substrate were used in both. Furthermore, coatings deposited using a powder obtained from one producer may not have the same characteristics as those deposited using a powder obtained from some other producer. This would be the case even if both powders have the same chemical designation and were deposited using the same process. Finally, coatings deposited by different processors should not be considered equivalent without comparative testing. This is the case even if the coatings were deposited using the same generic process and identical powders obtained from the same producer. The Boeing fatigue data seem to support this view; i.e., specimens HVOF coated by different processors can have different fatigue lives, in spite of the fact that the powders were procured to the same specification. It must always be remembered that WC-Co and WC-Co-Cr coatings are composite materials and, as such, their properties are governed not only by the carbide to metal ratio, but also by a host of other material and processing parameters.

3.2 The Importance of Adhesion & Bond Strength

The performance of any deposit, be that Cr plating or some HVOF coating, will depend not only on it being adherent to the substrate but also on the bond strength developed. For example, the effectivity of some deposit in corrosion or thermal/oxidation protection will depend, among other things, on the how long will it remain attached in place, as mechanical and thermal flexing take place. If this flexing generates stresses that exceed the bond strength, the deposit can get dislodged; a high bond strength would, therefore, be advantageous. A similar situation can arise in wear applications involving contact fatigue, if the hertzian stresses exceed the bond strength. Bond strength is also important in non-contact fatigue applications. In such applications, fatigue performance, is frequently rationalized in terms of residual stresses. Residual stresses develop as a result of constraint. Thus, in the absence of constraint, as would be the case for a non-adherent deposit, no residual stresses should form. Using the same rationale, it seems reasonable to conclude that, for adherent coatings, the magnitudes of the constraint and the resulting residual stresses would depend on bond strength.

Based on the above, it is clear that adhesion and bond strength are determining properties in the performance of deposits. In fact, it is thought that a reproducible performance of any deposit, in a given application, can not be ascertained unless the bond strength itself is reasonably reproducible. It is unfortunate, therefore, that adhesion and bond strength are often viewed as secondary properties. The issue is further compounded by the notion that adhesion is "something" that you can either have or not have and, as such, an exact determination of bond strength is not essential. Furthermore, bond strength and adhesion are often spoken of as one and the same. This, however, is not the case. Having 100% adhesion does not count for much if the bond strength is not satisfactory in the first place. Bond strength is not a generic "material property." Rather, it is a sensitive function of a host of material and processing parameters.

In the present Cr replacement efforts, bond strength determinations, when attempted, were performed per ASTM C 633 or its derivatives. As pointed out earlier, this test is limited by the strength of the adhesive used. With proper application, the deposits of interest here can develop bond strengths in excess of the capabilities of the commonly used adhesives. Thus, the otherwise quantitative test is rendered qualitative. This aside, porosity can introduce a problem when thermally sprayed coatings are tested per ASTM C 633. Adhesives with sufficient fluidity can penetrate through the pores and bond to the substrate, thereby altering test results. In such cases, the alternative would be to test after the coating had been sealed. Still, the adhesive-sealant bonds would alter the test results.

Evidently, bond strength testing based on ASTM C633 is inadequate for the task at hand. It is essential, therefore, that a truly quantitative test be adopted for determining the bond strengths of Cr plating and HVOF coatings. Any meaningful comparison between these deposits is not possible unless their respective bond strengths are known and are reasonably reproducible. The subject quantitative testing may be based on either the ring shear concept or on the Ollard method; Appendix A. Ring shear testing, however, can require the deposition of fairly thick coatings, which may not be representative of the thicknesses of interest. This would render the test inaccurate, since bond strength can depend on thickness. It is thought, therefore, that testing based on the Ollard method would be more suitable. Once a reproducible bond strength is determined, for a given shop practice, a qualitative bend test can be tailored for lot acceptance. The higher the bond strength the smaller would be the bend radius required to separate the deposit from the substrate, during bending.

3.3 The Origins of Conflict

Fatigue performance is a major concern in all the replacement efforts cited. In all cases, S-N data were generated, with the specific purpose of comparing Cr plating and HVOF coatings to each other and to the bare steel. There is general agreement that Cr plating involves a fatigue debit that becomes more substantial at lower stresses. Beyond this, the agreement ends. DOD, LBD and Airbus data indicate that HVOF coatings are associated with a fatigue debit that tends to increase at lower stresses. The Boeing data, by contrast, seem to indicate that HVOF coatings can give rise either to a fatigue debit or credit, depending on the stress level and on whether the specimens are notched or smooth. DOD data

Rev. A

show HVOF coatings to be superior to Cr plating; i.e., they involve lesser debits than Cr plating. Airbus data, by contrast, show essentially no difference; in fact, these data show Cr plating to be somewhat superior. To add to the confusion, the LBD data indicate that, for thinner coatings, especially at high stresses, Cr plating is generally superior to HVOF coatings, whereas for thicker coatings the trend is reversed. Evidently, there are inconsistencies, even conflict, in the fatigue results of the various efforts. In what follows, some potential culprits are explored. Although the emphasis here is on fatigue, some of the topics are equally applicable to other properties, such as wear performance and corrosion resistance.

3.3.1 Residual Stresses & HVOF Heat Input

The analyses presented in Appendix B strongly suggest that, whether an HVOF coating would involve a fatigue debit or credit and the magnitude thereof will depend on three factors. The first factor is the residual compressive stresses, resulting from powder peening. The second factor is the load carrying attributes of the coating; this factor becomes less significant as specimen size is increased. Both of these factors would favor fatigue credit. The third factor is the HVOF heat input, which would introduce adverse cooling residual stresses and could also degrade substrate strength. As such, this factor would favor fatigue debit. The magnitude of the fatigue debit/credit observed is the algebraic sum of the credits and debits resulting from these three factors. The Cr plating process, by contrast, will not affect heat-treated steel substrates, since it is performed at temperatures near ambient. Furthermore, the deposition process is always associated with adverse residual tensile stresses that far outweigh the beneficial load carrying attributes of the plating. As such, Cr plating will always be accompanied by a fatigue debit.

Thus, an HVOF coating would display a net fatigue debit if the heat-input effects prevail. As the heat input is increased, the magnitude of this debit will increase accordingly. In extreme cases, therefore, it is conceivable that the HVOF coating would involve more debit than Cr plating. The situation could be reversed, if the HVOF coating is deposited with lesser heat inputs. What is being said here is that the HVOF processing details will influence the coating performance, relative to that of properly applied Cr plating.

3.3.2 Adhesion & Bond Strength

In section 3.2, it is argued that residual stresses develop as a result of constraint. It is further argued that, for such constraint to materialize, the deposit must be sufficiently adherent; i.e., the bond strength achieved must be adequate. Thus, a deposit that ordinarily involves a fatigue debit, such as Cr plating, may show no debt at all if it is nonadherent. Accordingly, the fatigue performance of this nonadherent Cr plating would appear superior to that of a sufficiently adherent, debit-type HVOF coating, which, in turn, would appear superior to that of a sufficiently adherent Cr plating. It is felt that the true ranking of deposits with respect to each other and to the bare steel depends critically on the bond strength achieved. Put differently, the superior performance of Cr plating, relative to HVOF coatings, depicted in Airbus and LBD data could very well be due to adhesion problems in the former. Of course, a similar situation can arise solely from excessive heat inputs during the application of the HVOF coatings, as pointed out in section 3.3.1. What is being said here is that, a superior performance of Cr plating, relative to that of an HVOF coating, could be the result of plating and/or HVOF processing problems.

3.3.3 Thickness of the Deposit

The LBD data indicate that, as thickness is increased, performance of HVOF coatings will tend to improve, whereas that of Cr plating will tend to deteriorate. Furthermore, Appendix B makes the argument that, whether an HVOF coating would involve a fatigue debit or credit and the magnitude of that debit/credit will depend, among other things, on the thickness of that coating. It is conceivable that a culprit in data conflict could be differences in the thicknesses investigated in the various efforts. It is unfortunate that the only data published on the effects of deposit thickness are those of LBD. It is equally unfortunate that some publications do not list the thicknesses applicable to the fatigue curves shown, as if the data is generic to the deposit. Even when thickness is indicated, there is no mention of how that thickness was measured and its uniformity ascertained. There is more to the coating thickness issue; viz., oversize deposition.

Often, Cr plating and HVOF coatings are deposited oversize and then ground to the reported thickness. This technique can introduce its own uncertainties, since some characteristics of the thicker deposit will not simply be "erased" by grinding. For example, grinding is not likely to significantly alter the bond strength achieved by the thicker initial deposit. As such, the bond strength achieved by that thicker deposit would, for the most part, be retained after grinding. In the case of Cr plating, quantitative testing has shown that bond strength depends on thickness (Appendix A). Thus, after grinding, the thin plating would display the bond strength of the thicker initial deposit. Should bond strength influence the results, as it probably does (section 3.3.2), the data obtained would be interpreted as being those of the thinner deposit. This can lead to a discrepancy between what the data is thought to represent and what they actually represent. Such discrepancy would become more significant as the percent of the thickness removed by grinding is increased. In a similar fashion, bond strength of HVOF coatings is likely to also depend on thickness. Furthermore,

other characteristics are also expected to depend on thickness; viz., substrate heating effects (HVOF) and residual stress fields (Cr and HVOF). Each of the thickness dependent characteristics would generate a new set of uncertainties, if thicker deposits are applied and then ground to size. What is being said here is that the oversize deposition-grind technique could, at least in part, be a culprit in the inconsistent and conflicting data issue. Evidently, this technique needs to be controlled and its effects characterized.

3.3.4 Curve Fitting & Transitions in Fatigue Behavior

A possible culprit in the inconsistent and conflicting data issue could be the difficulties, and possible errors, in curve fitting, brought about by data scatter, typical of ultrahigh strength steels, and aggravated by the use of an insufficient number of specimens. This aside, the intersecting fatigue curves, depicted in LBD data, suggest an interesting potential culprit. Specifically, there are two types of intersections. In the first type, the fatigue curves for different deposits, at the same thickness, intersect. What this means is that a deposit that is superior to some other in the low cycle (high stress) regime becomes inferior at the high cycle (low stress) regime, and vice versa. In the second type, the fatigue curves of the same deposit, at various thicknesses, intersect. What this means is that, for a given deposit, one thickness that is superior to another in the low cycle regime becomes inferior at the high cycle regime, and vice versa. One possibility is that the intersections could be the result of the aforementioned data scatter and improper curve fitting. Another possibility is that the intersections could be due to the uncertainties brought about by the oversize deposition-grind technique, and by the inaccuracies in thickness determination (section 3.3.3). There is also the possibility that some of the intersections could be real, indicating actual transitions in fatigue behavior.

Transitions in fatigue behavior can be brought about by changes in the parameters or conditions that affect that behavior. For example, strength tends to control the high cycle (low stress) fatigue regime, whereas ductility tends to control the low cycle (high stress) regime. The change from one regime to the other, for a given substrate, is gradual and it does not occur at the same range for all deposits and thicknesses. Rather, the particular range where such change takes place varies from one deposit to the other and from one deposit thickness to the other, and it may also depend on the prevailing residual stress fields and on the specific crack initiation sites. [6]

It is important to note that a sufficient number of specimens need to be tested, in order to reveal the presence of a transition. Furthermore, the specimens should be strategically distributed throughout the entire S-N range of interest. Failure to take these measures may not allow the identification of transitions and, as such, can lead to grave errors in curve fitting. This may, at least in part, explain the inconsistencies in the data obtained in the various efforts.

4.0 COMMENTS & THOUGHTS

The following is offered with two goals in mind. First, to enhance awareness of certain aspects of Cr plating and HVOF coating performance. Second, to suggest means to improve confidence and minimize conflict in the data collected in ongoing and future efforts. The suggestions involve some research, suitable for academia and government laboratories. Ideally, this research should be collaborative in nature, through industrial consortia, with support from military and government agencies.

4.1 The HVOF coatings of interest here are composite materials. As such, their properties are governed not only by the carbide to metal ratio, but also by a host of other material and processing parameters. Therefore, material and process specifications for HVOF coatings should be prepared and controlled by the user, not by the powder producer or by the processor. Coatings deposited by different processors or powders procured from different producers should not be considered equivalent without comparative testing. Users should religiously enforce in-house receiving inspection for lot acceptance. Some of these precautions are also applicable to Cr plating.

4.2 Bond strength is neither a generic attribute nor is it a secondary property that may be ignored. Rather, it is a processing-sensitive property that depends on the strength, surface condition and pre-cleaning details of the substrate, the material and process specifics used in deposition, including deposit thickness, as well as on the details of post-deposition processing. Bond strength is a determining property that has significant influence on performance of the deposit. Therefore, all efforts should include quantitative tests for bond strength determination. A reproducible performance can not be ascertained unless the bond strength itself is reasonably reproducible.

4.3 The widely used ASTM C633 test and its derivatives are inadequate for bond strength determination for Cr plating and the HVOF coatings of interest here. Therefore, it is essential that a truly quantitative test be adopted for that purpose. It is thought that testing based on the Ollard method would be the most suitable. Once a reproducible bond strength is determined, for a given shop practice and substrate condition, then a qualitative bend test can be tailored for lot acceptance.

4.4 In as much as fatigue performance is involved, Cr plating should always involve a fatigue debit that becomes more substantial at lower stresses. HVOF coatings, on the other hand, may involve either a debit or a credit, depending on the processing parameters. As thickness is increased, fatigue performance of HVOF coatings should improve, while that of Cr plating should deteriorate. For these trends to be observed, the deposits need to be sufficiently adherent to the substrate. It appears that, for properly applied deposits of the same thickness, the HVOF coatings should display superior performance than that of Cr plating. When Cr plating displays a superior performance, this could be an indication of Cr plating and/or HVOF processing problems. Thus, the true ranking of Cr plating and HVOF coatings, with respect to each other and to the bare steel, would be influenced by the stress level and processing details. Processing details include the thickness at which the deposits are compared, the bond strengths achieved and, for HVOF coatings only, the heat input and powder grain size.

4.5 Tensile testing should be adopted as a quality assurance tool, in supplier qualification and lot acceptance, to detect gross substrate heating due to the HVOF process. Microhardness testing and metallographic examination should also be adopted to detect local heating effects just underneath the coating.

4.6 It is essential that accurate means be used to measure coating thickness and ascertain its uniformity. Furthermore, the oversize deposition-grind technique needs to be controlled and its effects characterized.

4.7 HVOF is a line-of-sight process. Due to this and other limitations, not all of the currently Cr plated components would be amenable to coating by that process. For such components, either the Cr plating has to be retained or some alternative to HVOF should be sought. Electroless nickel may be one such alternative.

Footnotes:

(1) Refer to: M.P. Nascimento, H.J.C. Voorwald, R.C. Souza and W.L. Pigatin, Plating & Surface Finishing, April 2001, pp 84-90.
(2) A joint USAF/GE study, in 1985, concluded that WC-12% Co coatings deposited by HVOF are similar to those deposited by D-Gun, in as much as bond strength, porosity and oxide content are involved.
(3) For D-Gun and HVOF coatings, respectively, 0.030 and 0.060 in. thick have been reported.
(4) Electroless Ni deposits are porous. The pores in thin deposits can expose the substrate, thereby adversely affecting corrosion resistance. A minimum of 1.5 mils (0.0015 in.) need to be deposited to ensure adequate corrosion protection. Post-deposition heat treatments (aging) increase the hardness of the deposit as well as bond strength, at the expense of corrosion resistance. The phosphorous (P) content governs the hardness and corrosion resistance of the deposit. In general, low P deposits have higher as-plated hardness and lower as-plated corrosion resistance than high P deposits. Electroless Ni deposits are associated with residual compressive stresses, with the low P deposits being the more prominent in that regard.
(5) Toxicity of electroless Ni plating solutions, however, is not as serious as of Cr plating solutions.
(6) In general, cracks may initiate at the coating, the coating-substrate interface, the substrate layer just underneath the coating or at the interior of the substrate, away from the coating.

APPENDIX A

ADHESION & BOND STRENGTH

A1.0 THE NATURE OF ADHESION

A1.1 Plated Deposits

A1.1.1 Electroplating:

It is generally believed that ele ctroplated deposits adhere to substrates via chemisorption, a form of short-range chemical interaction between atoms around the interface. This interaction is not believed to involve the formation of true chemical bonds (covalent, ionic and/or metallic). Rather, it is thought to involve the formation of associative intimate structures between coating and substrate atoms. The nature of these structures is not fully understood but they could involve atomic restructuring at the interface, Van der Waals forces and trapped substances (e.g., water, organic materials, air and/or ions). Micro-roughening the substrate surface increases the area of interaction, thereby increasing the force required to break the bond. Each plated layer would reproduce this micro r oughening and the subsequent layer would adhere via the same mechanism.

The need for roughened surfaces is thought to be the reason why the resulting interactions are often referred to as mechanical bonds. Be that what it may, the deposit and substrate atoms need to come in intimate contact, for any interaction to take place. Intimate contact would be brought about by the applied electrical field propelling the plating ions (+) towards the substrate (-). A prerequisite for intimate contact, however, is that substrate surfaces must be clean; i.e., free of grease, oil, oxides or any other substance that could interfere with achieving that intimate contact. This is achieved by degreasing, activation and similar processes. Underplates are often used to p revent substrate oxidation thereby enhancing adhesion of the top plate. Applying an immersion Zn underplate is the common practice in aluminum alloys; this is the zincate treatment. It is generally believed that this Zn underplate dissolves during subsequent plating operations, thereby exposing clean, unoxidized surfaces. In other cases, notably Ni underplates for Cr plating on steels, the underplate does not dissolve and it would act as a substrate for the desired plating.

Low melting deposits such as Zn, Pb and Sn-Pb can be fused after plating or they may be applied by hot dipping. In such cases, the deposited atoms diffuse into the substrate and true chemical bonds are formed. Unless adverse phases are formed as a result of this diffusion, bond strength is increased over that of as-plated deposit. These techniques can be extended to include higher melting deposits, provided that the heat required for melting does not adversely affect the substrate. Fusion and hot dip techniques are not applicable to Cr plating. An improvement in the bond strength of Cr, through diffusion, may be achieved by heating above about 1600 F. This elevated temperature heating, however, is likely to have adverse effects on most common substrates.

A1.1.2 Electroless Plating:

Electroless plating is performed by chemical means, without the use of an external electrical field. There are many ways in which electroless plating can be performed, depending on the coating material and the equipment available. In this work, however, will focus on autocatalytic electroless Ni (EN), a competitor to Cr plating in many applications. As in electroplating, adhesion of electroless Ni is believed to be largely mechanical. In electroless plating, however, there are no external electrical fields to propel the plated ions towards the substrate. Tentatively, then, one would expect bonding of electroless deposits, to any given substrate, to be generally less strong than those of their electroplated counterparts. Direct comparisons, however, are not readily available in the literature. What is available is an abundance of statements, without supporting details or numbers, indicating that EN plating, applied to steel and aluminum, develop higher bond strengths than those attained by Cr plating on the same substrates. These statements led some to theorize that, in addition to mechanical bonding, some chemical bonds must be involved in as-plated electroless deposits. While this may or may not be the case, it is an established fact that post-plate heating leads to Ni diffusion into the substrate. This, in turn, leads to the formation of true chemical bonds with the attendant increase in bond strength; depending on the substrate, the required temperatures are in the 240-750 F range.

A1.2 Thermally Sprayed Deposits (Coatings)

Adhesion of thermally sprayed coatings is also believed to be via the formation of mechanical bonds. Micro roughening and cleaning here are as important as they are in plating. The force by which the molten or partially molten droplets impact the substrate surface is a factor that aids in achieving intimate contact. Thus, the higher the speed by which these droplets are propelled towards the substrate, the more likely that a stronger bond can be achieved. That is why HVOF, D-Gun and SDG coatings have some of the highest bond strengths among all thermal spray coatings. This aside, the elevated temperatures, characteristic of thermal spray, are expected to lead to some short range diffusion of the coating atoms into the substrate and the attendant formation of chemical bonds. More diffusion and chemical bonding would be expected by heating, subsequent to coating, at an appropriate temperature, with or without fusion. Unless adverse phases are formed as a result of diffusion, bond strength would be increased over that of the as-sprayed coatings. As is always the case, the selected heating temperature should not adversely affect the substrate.

A1.3 A Note of Caution

Often, bond strength and adhesion are spoken of as one and the same. This, however, is not the case. Having 100% adhesion does not count for much if the bond strength is not satisfactory in the first place. Bond strength is not a "material property." Rather, it is a sensitive function of a host of material and process parameters. A lower bound may be set by requiring the bond strength to equal or exceed the cohesive strength of the weaker of the substrate and deposit. If this cannot be achieved by current / best shop practices for some deposit, a different type of deposit should be sought, provided that the new deposit satisfies all other service requirements. With

the difference between adhesion and bond strength in mind, the two terms will be used interchangeably in what follows, unless a distinction is required.

A2.0 THE IMPORTANCE OF ADHESION & BOND STRENGTH

The performance of any deposit will depend not only on it being adherent to the substrate but also on the bond strength developed. For example, the effectivity of some deposit in corrosion or thermal / oxidation protection will depend, among other things, on the how long that deposit will remain attached in place, as mechanical and thermal flexing take place. If this flexing generates interface stresses that exceed the bonding strength, the deposit can get dislodged; a high bond strength would, therefore, be advantageous. A similar situation can arise in wear applications involving contact fatigue, if the interface hertzian stresses exceed the bond strength. Bond strength is also important in non-contact fatigue applications. In such applications, fatigue performance, is frequently rationalized in terms of residual stresses. Residual stresses develop as a result of constraint. Thus, in the absence of constraint, as would be the case for a non-adherent deposit, no residual stresses should form. For adherent deposits, it seems reasonable to conclude that the magnitudes of the constraint and the resulting residual stresses would depend on bond strength; note that the residual stresses developed will tend to affect the apparent bond strength that will be actually measured. It is clear, therefore, that adhesion and bond strength are determining properties in the performance of any deposit. In fact, it is thought that a reproducible performance, in any given application and on any given substrate, can not be ascertained unless the bond strength itself is reasonably reproducible. Unfortunately, however, adhesion and bond strength are often viewed as secondary properties. Some believe that adhesion is "something" that you can either have or not have and, as such, an exact determination of bond strength is not essential.

A3.0 ADHESION VERIFICATION

A3.1 General

Plating and thermal spray specifications generally require some form of testing to verify adhesion of the deposit. These tests may or may not involve actual determination of bond strength. The tests generally originated in conjunction with plated metallic deposits. Adhesion tests are classified in two groups; namely, qualitative and quantitative. Qualitative testing is intended to verify that there is adhesion, not to provide information on actual bond strength. Quantitative testing, by contrast, is intended to provide specific information on actual bond strength.

Qualitative adhesion tests include, but are not limited to, the bend test, the burnishing test, the chisel-knife test, the draw test, the file test, the grind-saw test, the heat-quench test, the impact test, the peel test (by tape or solder), the push test and the scribe-grind test. There is no national standard, accepted industry-wide, that defines the details of such tests for all types of deposits. One set of definitions for the aforementioned tests, applicable to metallic deposits, is given in ASTM B 571, to be discussed shortly. Variants of these tests, custom tailored to suit a specific clientele, do exist and are being used; e.g., the knife-peel test of ANSI/AWS C2.18 and the tape tests of ASTM D 3359.

Tests for quantitative measurement of adhesion can be divided into three groups. [1] The first group utilizes solders, adhesives, or thick deposits to attach some form of grip to the deposit through which a detaching force can be applied. The tests provide either tensile or shear strength information. An example of the first group is ASTM C 633, used for thermal spray coatings, where an adhesive is used to determine tensile bond strength. A variant of this test, ASTM D 4541, employs portable testers available through specific suppliers. The second group includes tests that directly determine tensile bond strength, without the use of adhesives, solders or any other means to attach a grip. These tests are generally based on the Ollard method, [2] its modifications [3-5] and a variant. [6] The third group includes a variety of specialized techniques in which the primary mode of stress is shear, and do not employ adhesives, solders or any other means to attach a grip. The most popular of these is the ring shear test. [7-10]

A3.2 Adhesion Verification in the Plating & Coating Industries

A3.2.1 Plated Deposits

Specifications for plating of metallic deposits generally require some qualitative form of testing for adhesion verification, with or without reference to ASTM B 571. ASTM plating specifications generally call out the specific tests of ASTM B 571. For example, ASTM B 650 (Cr plating) requires adhesion to be verified by the bending, filing, heat-quench or push tests of ASTM B 571. By contrast, the popular AMS QQ-C-320 (formerly, QQ-C-320) Cr plating specification spells out the details of the chisel-knife and bending tests required for adhesion verification, without referencing ASTM B 571; the bend test is a variant of that in the ASTM. The current revision of AMS 2406, another Cr plating specification, requires adhesion verification per ASTM B 571, but does not specify the particular test(s). Earlier revisions, however, made no reference to ASTM B 571 and left adhesion testing to be as agreed upon between purchaser and vendor. Mil-C-26074, for EN plating, references ASTM B 571 and describes a modified bend test with added requirements.

A3.2.2 Thermally Sprayed Coatings

A3.2.2.1 General: Specifications for thermal spray coatings also vary in their requirements for adhesion verification. Many specifications, however, require both qualitative and quantitative testing for this verification. Bending is the most widely specified form of qualitative testing. Bend test details are typically listed in the specification without reference to ASTM B 571. Quantitative adhesion verification is almost exclusively through the tension test of ASTM C 633. This ASTM specification is actually for flame sprayed coatings, but has been adopted by the technical community for all thermally sprayed coatings. The test has even been utilized to verify adhesion of Cr plating in some of the present Cr replacement efforts.

A3.2.2.2 The Trouble with ASTM C 633: In the ASTM C 633 test, an adhesive is used to bond a plated/coated specimen of the substrate material to a loading extension. The bonded assembly is then tested, in tension, to failure. Failure can occur in the plating/coating, the

substrate, the plating/coating-substrate interface or the adhesive. The latter failure could be cohesive, within the adhesive itself, or adhesive, at either bondline; one bondline is at the plated/coated substrate and the other at the loading fixture. For the purposes of this work, both adhesive and cohesive failures of the adhesive will be referred to as "failure of the adhesive." The cohesive strengths of the WC coatings of interest here and those of most plated deposits, including Cr, are large enough to preclude coating failure. The same is true for s ubstrates made of steel and most other metallic alloys. Thus, failure would be expected at the adhesive or at the plating/coating-substrate interface. Bond strength of the plating or coating can only be determined when failure occurs at that interface. This is why ASTM C 633 requires that the strength of the adhesive be greater than the bond strength being determined. This is not a problem for most thermal spray coatings, where the bond strength is in the 3-10 ksi range; the practical limit for organic adhesives is around 10 ksi. For D-Gun, SDG, HVOF and some plasma coatings, however, this may not be the case. When properly applied, coatings sprayed by these methods can develop bond strengths far in excess of 10 ksi. The same is true for Cr plating. In such cases, the otherwise quantitative test is rendered qualitative. This aside, porosity can introduce a problem when thermally sprayed coatings are tested per ASTM C 633. Adhesives with sufficient fluidity can penetrate through the pores and bond to the substrate, thereby altering test results. In such cases, the alternative would be to test after the coating had been sealed. Still, the adhesive - sealant bonds would alter the test results.

A3.3 Laboratory Techniques

Direct tension and shear test methods, [1-10] where no external means of attaching a grip are employed, are largely laboratory tools for investigating bond strength. These methods are generally not commercially attractive, due to the elaborate set-ups required and the time consuming nature of the tests. In this section, some of the results obtained for the coatings of interest here will be presented. As will become evident shortly, the results clearly show that bond strength is not a generic "material property." Rather, it is a processing-sensitive property that depends on the strength, surface condition and pre-cleaning details of the substrate and on the material and process specifics used in deposition, including deposit thickness. The results further show that the bond strengths developed can be far in excess of the capabilities of known organic adhesives, of the types used in ASTM C 633.

A3.3.1 Cr Plating: Zmihorski, [7] using the ring shear test, investigated the effects of several variables on the bond strength of har d Cr plating to steel (1.1% C, 1% Mn, 1.2 Cr, 1.5 % W), heat-treated to HRC 60. He found out that thinner deposits have higher bond strengths and that bond strength depended critically on cleaning prior to plating. Higher current densities resulted in lower bond strengths; he attributes this to the more substantial H pick-up at the higher current densities. Plating bath contamination with Fe reduced bond strength. In general, lowering the strength of the substrate increased bond strength; he warns, however, that lowering substrate strength may not be suitable for contact fatigue situations. The shear bond strengths he reported, for a wide range of cleaning methods, plating variables and coating thicknesses, ranged from about 50 ksi to lower than 15 ksi. He pointed out, however, that under the then best shop practices, shear bond strengths in the 28-32 ksi could be obtained. This translates to tensile bond strengths in the 42-48 ksi (using the formula $F_{su} = 0.66 F_{tu}$). Clearly, such strength levels exceed that of the best available organic adhesive, of the types used in ASTM C 633.

A3.3.2 EN Plating: There is a host of promotional EN publications that quote bond strength values but fail to indicate whether these values are shear or tensile and, in some cases, without indicating the substrate. As such, these quotations are ignored here. This aside, Dini and Johnson, [9] using the ring shear test, determined the bond strength of EN to AISI 4340 and HP-9-4-20 steels; the authors, however, did not indicate the strength levels of the steels. The steels were cleaned and then HCl pickled. The reported shear bond strengths were about 55 and 1.5 ksi for the AISI 4340 and HP-9-4-20, respectively. When a sulfamate Ni underplate was used on AISI 4340, the reported shear bond strength was increased to about 67 ksi. A wood's Ni strike, applied to HP-9-4-20 prior to EN plating, increased the shear bond strength to about 15 ksi. The best result obtained, about 75 ksi, was obtained when HP-9-4-20 was anodically cleaned in H_2SO_4 with a subsequent Wood's Ni strike, prior to EN plating. Thus, the available data indicate that shear bond strengths in excess of 15 ksi can be achieved with EN plating on steels; the corresponding tensile bond strength is 23 ksi, using the formula $F_{su} = 0.66 F_{tu}$. Again, such strength level exceeds that of the best available organic adhesive, of the types used in ASTM C 633.

A3.3.3 WC-Co Coatings : In some old Union Carbide publications, it is stated that, for the WC-9Co (LW-1), WC-15 Co (LW-1N40), WC-13 Co (LW-1N30) and WC-10Co-4 Cr (LW-15) coatings, applied by the D-Gun, the bond strengths exceed the strength of the adhesive used in testing per ASTM C 633. It is further stated that for the first two coatings, special laboratory methods revealed bond strength values in excess of 25 ksi. The publications, however, did not indicate any test or substrate specifics and did not identify whether the 25 ksi value was in shear or tension. As was the case for Cr and EN, such strength level exceeds those of the known organic adhesive, of the types used in ASTM C 633.

A4.0 COMPARING COATINGS

A4.1 Historical

For the past three decades or so, EN has been a serious competitor to engineering (hard) Cr in wear and corrosion applications, as well as in salvage operations. As indicated above, it is thought that bond strength is a determining factor in the performance of any deposit. In spite of that, there appears to be a certain lack of quantitative data comparing the bond strength of EN to that of Cr, deposited using the best shop practices of the time, on the same substrate types. Only statements as to superiority of EN are published, without supporting details or numbers; these statements are basically in publications promoting EN. Now, a group of new contenders, WC coatings, enters. In as much as bond strength is involved, these coatings are compared, and frequently deemed "equivalent," to Cr via the ASTM C 633 test. A review of the bond strength values presented in the preceding section indicates that the bond strengths of Cr, and certain WC coatings can exceed the strength of any of the existing organic adhesives, of the types used in the ASTM test. Thus, the test is inadequate for comparing the subject coatings.

<u>**A4.2 The Need for Quantitative Adhesion Measurements**</u>

It is evident from the preceding sections that adhesion verification, for the deposits of interest here, has been qualitative in nature. Specifically, adhesion verification in plated deposits is through bend testing and similar qualitative methods. For WC coatings applied properly by HVOF, SDG and D-Gun, bond strengths typically exceed the strength of the adhesive and, as such, the otherwise quantitative test is rendered qualitative. This is the state of the art at present. It is felt, however, that the effect of any deposit on the properties of a given substrate can not be reproduced unless the bond strength itself is reasonably reproducible. The same holds true for any meaningful comparison between deposits. Only quantitative bond strength measurements can provide the required information. What is being said here is that the deposits of interest in the Cr replacement efforts cannot be fully characterized or compared without quantitative measurement of their bond strengths. It is essential, therefore, that a truly quantitative test be adopted for determining the bond strengths of Cr plating and HVOF coatings. The subject quantitative test may be based on either the ring shear concept or on the Ollard method. Ring shear testing, however, often requires fairly thick deposits, which may not be representative of the thicknesses of interest. This would render the test inaccurate, since bond strength depends on thickness (section A3.3). It is thought, therefore, that some test based on the Ollard method would be more suitable. Once a reproducible bond strength is determined, for a given shop practice and substrate condition, a qualitative bend test can be tailored for lot acceptance. The higher the bond strength, the smaller would be the bend radius required to separate the coating from the substrate, during bending.

Appendix A References:

(1) D. Davies and J. A. Whittaker, Metallurgical Reviews, vol. 12, 1967, pp 15-26.
(2) E. A. Ollard, Met. Trans. Far. Soc. 21, 1925, pp 81-90.
(3) A. W. Hothersalt, Trans. Electrochem. Soc. 64, 1933, p 69.
(4) E. J. Roel, Iron Age 146, Sept. 26, 1940, pp 17-20.
(5) E. J. Roel, Iron Age 146, Oct. 3, 1940, pp 30-33.
(6) B. B. Knap, Metal Finishing, 47, Dec. 1949, p 42.
(7) E. Zmihorski, J. Electrodepositors' Tech. Soc., vol. 23, 1947-48, p 203.
(8) J. W. Dini, J. R. Helms and H. R. Johnson, Electroplating and Metall Finishing, March 1972, p5.
(9) J. W. Dini and H. R. Johnson, presented in Electroless Nickel Conference III, Chicago, 29-31 March 1983.
(10) STP 640, ASTM.

APPENDIX B

HARD CHROME REPLACEMENT EFFORTS

Significant efforts are underway to evaluate the merits of HVOF WC coatings in comparison to hard (engineering) Cr plating. These efforts include those by the Department of Defense (DOD), Boeing-Long Beach Division (LBD), Boeing-Seattle (Boeing) and Airbus Industries. In what follows, brief descriptions of these efforts and the available results is presented. In what follows, hard (engineering) Cr plating will simply be referred to as Cr plating.

B1.0 THE DOD EFFORTS

<u>B.1.1 The HCAT Effort</u> [1,2]

In the early 1990's, several DOD sponsored studies compared the properties of different HVOF coatings with those of Cr plating; e.g., the DARPA sponsored study by Northwestern University. While the studies generally revealed that some HVOF coatings possessed certain desirable attributes, no technology insertion efforts were attempted. In 1996, DOD initiated an effort to demonstrate and validate HVOF coatings as viable replacements to Cr plating on aircraft components. The effort was executed by the Hard Chrome Alternatives Team (HCAT), a tri-service/industry group that was formed in 1996. HCAT includes DOD research laboratories, aircraft depots, aircraft OEM's, materials testing laboratories, an HVOF equipment manufacturer and an FAA-certified repair facility that has both Cr plating and thermal spray capabilities.

<u>B1.1.1 Phase 1, Initial Tasks</u>

The initial tasks involved conducting coupon fatigue, corrosion, hardness and wear testing. The coupons (substrates) were made of AISI 4340 (F_{tu}: 260-280 ksi) steel, 7075 aluminum alloy and PH13-8Mo stainless steel. The coupons were either Cr plated or HVOF coated. Two HVOF coatings were selected for evaluation; namely, WC-17/18% Co and Tribaloy 400, a CO-Mo-Cr alloy. For the purposes of this appendix, will focus on the AISI 4340 steel.

Figure 1: DOD Axial Fatigue Trends. Schematic

<u>(a) Fatigue</u> : S-N test results indicate that Cr plating leads to a substantial fatigue debit, compared to uncoated steel. The HVOF coatings involved a much less pronounced fatigue debit. For all coatings, the debit appeared to increase at lower stress levels. These trends are shown schematically in Figure 1; the applicable plating/coating thickness, however, was not listed.

<u>(b) Corrosion</u>: Q-Fog cabinet corrosion tests (ASTM B117) indicated virtually no difference between the HVOF coatings and Cr plating. In atmospheric corrosion testing, however, the HVOF WC-17% Co coating performed markedly better than hard Cr or HVOF Tribaloy 400.

<u>(c) Hardness</u>: Microhardness testing showed the WC coating to be somewhat harder than the hard Cr. The Tribaloy hardness was significantly lower.

<u>(d) Wear</u> : Abrasive wear testing revealed that the HVOF WC-17% Co is superior to hard Cr plating, which in turn is superior to HVOF Tribaloy 400. These results suggest that abrasive wear is influenced by hardness.

<u>B1.1.2 Phase 2, Validation</u>

The above work demonstrated the technical viability of HVOF coatings in replacing Cr plating. Full-scale production validation required significant additional work. It was realized that the key to full qualification is the involvement of all stakeholders; that is to say, repair depot technical representatives, cognizant engineering activities from the OEM's as well as the services and weapons system / single item program managers. It was further realized that different stakeholders would be involved in the qualification of each aircraft component. The HCAT joined with two other DOD working groups to execute separate projects to fully qualify HVOF coatings on the following classes of aircraft components: (a) landing gear inner cylinders, axles, pins and actuators (in conjunction with Boeing, and the landing gear manufacturers, BF Goodrich/Menasco, Heroux and Messier-Dowty); (b) propeller hubs (in conjunction with Hamilton Standard); (c) hydraulic actuators for systems other than landing gear (OEM's to be identified); (d) helicopter dynamic components, including transmission and rotor head components (in conjunction with Boeing Philadelphia and Sikorsky) and; (e) gas turbine engine (GTE) parts (in conjunction with GEAE, P & W and RR Allison). The GTE project uses basis materials that differ from those of the other projects, includes five coatings in addition to WC-17% Co and investigates both HVOF and the plasma spray methods; see reference 3.

The execution plan for each project called for the development of test protocols, which define all of the necessary testing required for qualification. These generally included coupon testing for fatigue, corrosion and wear of the selected HVOF coatings on base materials appropriate for the type of component being considered. The protocols also called for actual component testing on test rigs and on operational aircraft where performance may be tracked over extended periods of time. Apart from developing and executing the test protocols, the projects also involved (a) establishing production-level HVOF capabilities at the appropriate maintenance facilities, (b) personnel training, (c) cost analysis and (d) developing standards for application, grinding and stripping the various components. The entire landing gear test protocol is available on the web (www.hcat.org).

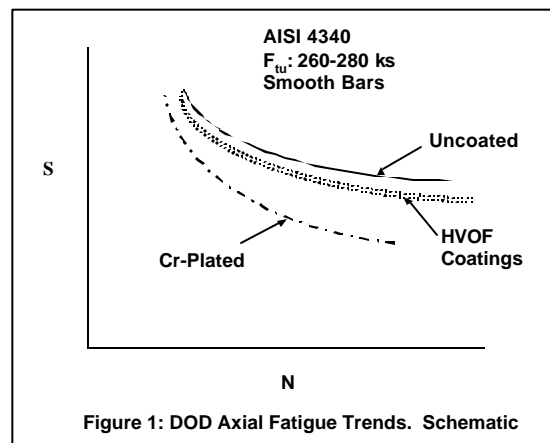The most developed of the projects is that of the landing gear, which, as of 1997, became a joint effort between the US and Canada. A Canadian team, C-HCAT, was formed from the Department of National Defense and Industry Canada to execute some aspects of the testing; interestingly, Canadian companies manufacture more than two thirds of the landing gear on military and commercial aircraft in North America. Many of the US and Canadian team members are also involved in the Boeing, LBD and Airbus efforts, to be described later on. The landing gear project involves coupon and component testing. Fatigue coupon testing is underway, using HVOF WC-17% Co (the US team) and HVOF WC-10% C0-4% Cr (the Canadian team) deposited on peened and unpeened AISI 4340, 300M and AerMet 100 steels. Initial results confirm the superiority of HVOF coatings over Cr plating and also indicate a peening benefit in all coatings, as depicted schematically for AISI 4340 in Figure 2; apparently, no data were available for 300M or AerMet 100 at the time of publication.

Although no strength range is listed in the reference, it is believed to be the same as that in Figure 1. It appears that both Cr plating and the HVOF coating involve a fatigue debit, compared to the uncoated steel. Rig testing at Boeing St. Louis involves F-18E/F main landing gear with several HVOF coated components. Flight-testing is on Navy P-3 and E-6A aircraft that are currently in service.

Figure 2: DOD Axial Fatigue Trends. Landing Gear Coupons. Schematic.

B1.2 The Naval Air Warfare Center (NAWC) Effort [4]

NAWC is utilizing AerMet 100, a landing gear steel. AerMet 100, developed by Carpenter, is an ultrahigh strength steel, hardenable by both the martensite and precipitation reactions; nominal alloy content is 13.4% Co, 11.1% Ni, 3.1% Cr, 1.2% Mo, 0.23% C. AerMet specimens were either Cr plated or HVOF coated with WC-17% Co, and the thickness investigated was 3 mils (.003 in.). The HVOF coating was compared to Cr plating under stress corrosion, corrosion fatigue and fatigue conditions.

(a) Stress Corrosion: Cr plating was superior to the HVOF coating. This is because the latter tended to spall and expose the substrate, as result of the tensile loading associated with stress corrosion testing.

(b) Corrosion Fatigue: The HVOF coating was superior to Cr plating which, in turn, was superior to the bare steel.

(c) Fatigue: S-N data indicate that the HVOF coating generally involves less fatigue debit than Cr plating. The debit associated with either appeared to increase at lower stresses.
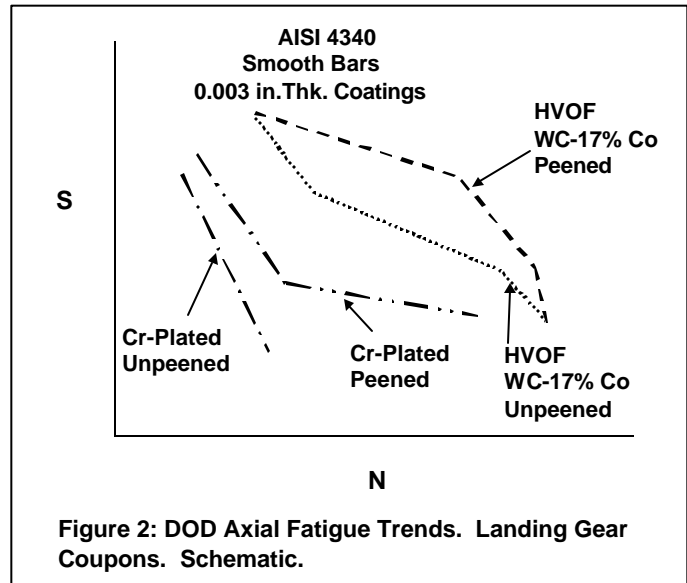
B2.0 THE LBD EFFORT

During the 1990's, a major effort was launched at LBD to compare Cr plating and WC-17% Co, applied by the HVOF, D-Gun and SDG thermal spray processes. A fairly detailed account of that effort is given below.

B2.1 Phase 1, Laboratory Testing [5]

B2.1.1 Corrosion Testing: Test panels were made of AISI 4130 steel, WC coated or Cr plated. The WC coating was applied using the methods described above. Corrosion testing was per ASTM B117. Testing did not reveal any major differences between the Cr plating and the thermally sprayed coatings. This is in agreement with the results of DOD-HCAT testing, per the same ASTM, where the HVOF coating and Cr plating were deemed equivalent.

B2.1.2 Hydrogen Embrittlement Testing: The purpose of hydrogen embrittlement testing was to determine whether the strippers used for stripping the WC coating would lead to hydrogen degradation. Heat-treated AISI 4340 steel substrates were coated by SDG and HVOF and then electrolytically stripped. This was followed by hydrogen embrittlement testing per ASTM F519. Apart from having to retest one specimen, no anomalies were reported.

B2.1.3 Metallography: The purpose of metallographic testing was to examine the microstructure of the WC coating as applied, on AISI 4130 specimens, using HVOF, D-Gun and SDG; Cr plating was not included. Two specimens per application process were used. Metallographic examination was performed per the requirements of the applicable company specifications. There were two cases (both HVOF) of excessive residual carbide content, three cases (1 D-Gun and 2 SDG) of hardness below requirement, one case (HVOF) of interface separation, four cases (1 D-Gun, 1 SDG and 2 HVOF) of coating cracks; these cracks, however, were suspected to have been caused by substrate flexing. Certainly, these results cast some doubt on coating specimen preparation, on pre-coating surface cleaning, on the particulars of some of the spray techniques used and on handling for metallographic examination.

B2.1.4 Wear Testing: Wear testing was performed using the block on ring method (ASTM D3704), with the blocks Cr plated or WC coated, using HVOF, D-Gun or SDG. The subject test method, however, is really intended to assess the wear preventative qualities of lubricating greases, not for coating and/or plating comparisons. Furthermore, the actual tests involved determinations of weight loss, which is not required by the ASTM. Finally, the block and ring materials used were neither those specified in the ASTM nor did they represent materials that are used in actual hardware. This aside, the results suggest that sprayed WC is superior to Cr plating. It should be noted that flaking of the coating due, for example, to poor adhesion, would be reported as weight loss. This may have been a culprit in the data obtained.

B2.1.5 Bond Testing: Bond strength testing was performed in a manner that is similar to that of ASTM C633. In this test, a specimen is plated/coated and then adhesively bonded to another bare specimen. The bonded assembly is then tested in tension, to failure. The LBD tests were performed using AISI 4130 steel specimens, with the plating/coating applied to the bare specimen or over an undercoating of sulfamate Ni; coating was by HVOF, D-Gun or SDG. ASTM C633 type tests are really intended for thermal spray coatings with bond strengths that are less than the capability of the adhesive. When bond strength of the coating/plating exceeds the capability of the adhesive, the test is rendered qualitative in nature. According to the controlling company thermal spray specifications, a coating is acceptable if its bond strength exceeds 10 ksi. Strictly speaking, the 10 ksi requirement, extracted from thermal spray specifications, should not be applicable to plated Cr. This aside, the results obtained ranged from a few hundred psi to around 13 ksi. Strength values in excess of 10 ksi could be real, representing adhesive failures at the practical limit. Lower values could be due to poor coating adhesion or improper test set up, with the later including improper adhesive curing, unleveled bonded surfaces, misalignment of bonded specimens and the like. The results, however, were dismissed as being the result of improper test set up.

B2.1.6 Bend Testing: The bend test was performed using specimens made from AISI 4130 sheet stock. The Cr plating and WC coating were deposited either directly onto the specimens or over a sulfamate Ni under coating; coating was by HVOF, D-Gun or SDG. Upon bending, all deposits cracked, which is acceptable according to the controlling specifications. For WC coatings deposited on sulfamate Ni, the cracks penetrated into the substrate. Spalling occurred in all cases where Cr was plated directly onto the specimens. This spalling was attributed to poor adhesion, resulting from improper cleaning practices.

B2.1.7 Tensile & Fatigue Testing: The tensile and fatigue specimens were either 300 M steel, heat treated to 280-300 ksi F_{tu}, or Hy-Tuf steel, heat treated to 220-250 ksi F_{tu}. All the tensile and fatigue specimens made from each steel were heat treated in the same lot; decarburization control measures were invoked. After heat treatment, the specimens were finish machined, nital etched and then baked for hydrogen embrittlement relief. This was followed by cleaning and Cr plating/WC coating, directly onto the specimens; no shot peening was used. The WC coatings were applied by HVOF, D-Gun or SDG. The plating/coating thicknesses investigated were 0.003 and 0.012 in., for the tensile specimens, and 0.003, 0.006 and 0.012 in., for the fatigue specimens. The Cr plating and WC coating were deposited oversize and then ground to the reported thicknesses. It is not clear, however, as to what method was used to measure coating thickness and assertion its uniformity. Some specimens were not tested bare, as control specimens, to establish a base line for each steel.

B2.1.7.1 Tensile Testing: Tensile testing was performed per ASTM E8. The results indicated that the control, Cr plated, D-Gun and SDG specimens were all within the accepted range for both 300M and Hy-Tuf. By contrast, all HVOF coated 300M specimens and one HVOF coated (0.012 in. thick) Hy-Tuf specimen fell below the acceptance range. Furthermore, the observed tensile strength degradation was generally more pronounced for thicker coatings. These results indicate that softening (tempering) of the base metal took place as a result of the HVOF application of the coatings. Indeed, experimentation revealed that the culprit is the fast deposition rates employed. The implication here is that the high heat inputs, associated with fast deposition rates, must have led to the observed degradation. It is interesting to note that, while HVOF processing degraded all 300M specimens, it only degraded one Hy-Tuf specimen. This may be due to utilizing higher heat inputs in processing 300M. Alternately, it may just be that 300M, in the 280-300 ksi strength range, is more heat sensitive than Hy-Tuf, in the 220-250 ksi strength range. In other words, it may just be that, at their respective strength ranges, 300M has a steeper tempering curve than that of Hy-Tuf; note that both steels are ordinarily tempered in the 500-600 F range to develop their specified strength ranges.

B2.1.7.2 Fatigue Testing: S-N data were obtained by testing per ASTM E466. Test frequency was 30 Hz maximum until failure or 12,000,000 cycles, whichever came first. Data scatter was pronounced, as would be expected from the high strength steels being investigated; shot peening would have reduced the scatter. Schematic representations of the reported results for 300M and Hy-Tuf are shown, respectively, in Figures 3 and 4. While the original publication made no mention of any trends, an attempt at their identification is made here. In discerning such trends from the data, one must be cognizant of the difficulties, and possible errors, in curve fitting, brought about by a number of factors. First is the data scatter, typical of unultrahigh strength steels. Second is fact that the specimens were not shot peened, a condition that tends to accentuate data scatter. Third is the use of an insufficient number of specimens. Fourth is the fact that, in some cases, there is evidence of improper processing, prior to and/or during coating application. Having said this, it should be remembered that, while fatigue performance is important, it may not be the overriding concern in landing gear applications. This is because rework is typically required long before normal fatigue life had been exhausted.

(a) In general, SDG coatings displayed superior fatigue performance than all the other deposits. For Hy-Tuf, there was always an SDG fatigue credit that generally increased with increasing coating thickness. In the case of 300M, thinner SDG coatings displayed a fatigue life debit that tended to decrease with increasing coating thickness, eventually turning into a credit for the thickest coatings. HVOF coatings on either steel displayed a similar trend. Specifically, there was always an HVOF fatigue life debit that, for the most part, tended to decrease with increasing coating thickness; a fatigue debit was also reported for HVOF coatings in the DOD efforts. Thus, it is reasonable to conclude that, as the HVOF or SDG coating thickness is increased, fatigue life debit is reduced, and may eventually be replaced with a fatigue credit. The data also indicate that, as the applied stress is changed, at some given coating thickness, the fatigue life debit/credit changes accordingly. The general trend is that a decrease in the applied stress is accompanied by an increase in debit or a decrease in credit, as applicable. This trend is even evident in the case of the 300M HVOF coating, where the heat input had degraded the strength of the substrate. A notable deviation is the case of the 300M SDG coating, where the opposite trend is depicted. Other, less notable, deviations exist. These deviations, however, are thought to be due to curve fitting problems.
(b) In contrast to SDG coatings, D-Gun coatings generally displayed the worst fatigue performance of all deposits. This is a rather surprising result, since both coating processes were developed by, and applied at, the same facility.
(c) Cr plating was associated with a fatigue debit. The debit tended to increase with coating thickness. This progressive deterioration with thickness is clearly evident in the Cr-plated Hy-Tuf data (top right corner of Figure 4). The reason for this is that Cr electrodeposits are always accompanied by residual tensile stresses. The magnitudes of these stresses tend to increase as the plating thickness is increased, leading to a corresponding deterioration in fatigue performance; i.e., a corresponding increase in fatigue debit. For such a deterioration trend to be observed, however, the Cr plating in all specimens evaluated must be sufficiently adherent to the substrate; otherwise the trend could be masked, even reversed. A reversal in the trend is evident in the case of 300M (top right corner of Figure 3), where the 0.006 in. thick plating displays a larger debit than the 0.012 in. thick plating. This seems to suggest that the 0.012 in. plating
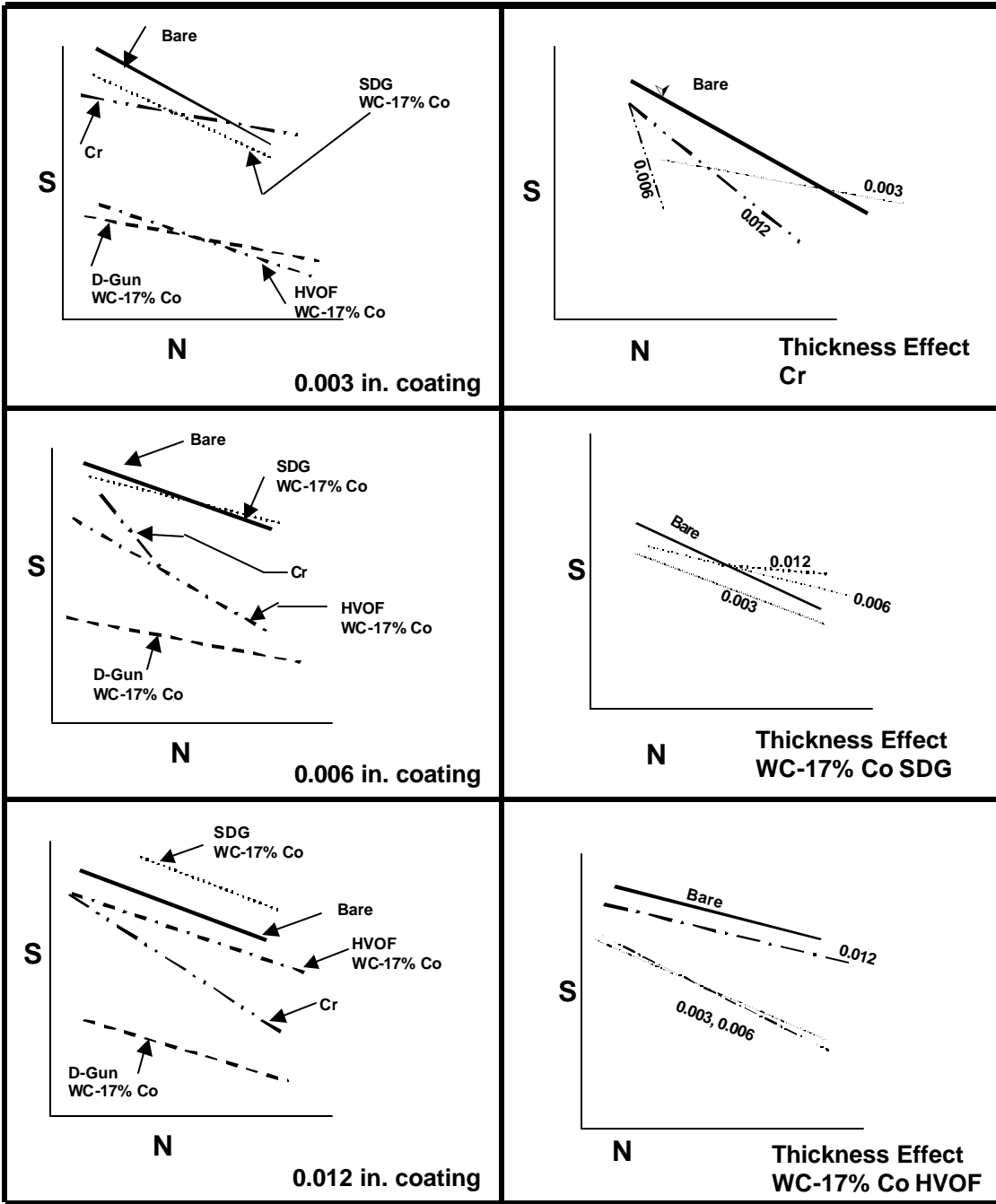
Figure 3: LBD Axial Fatigue Data. 300M (F_{tu}: 280-300 ksi). Smooth Bars. Schematic
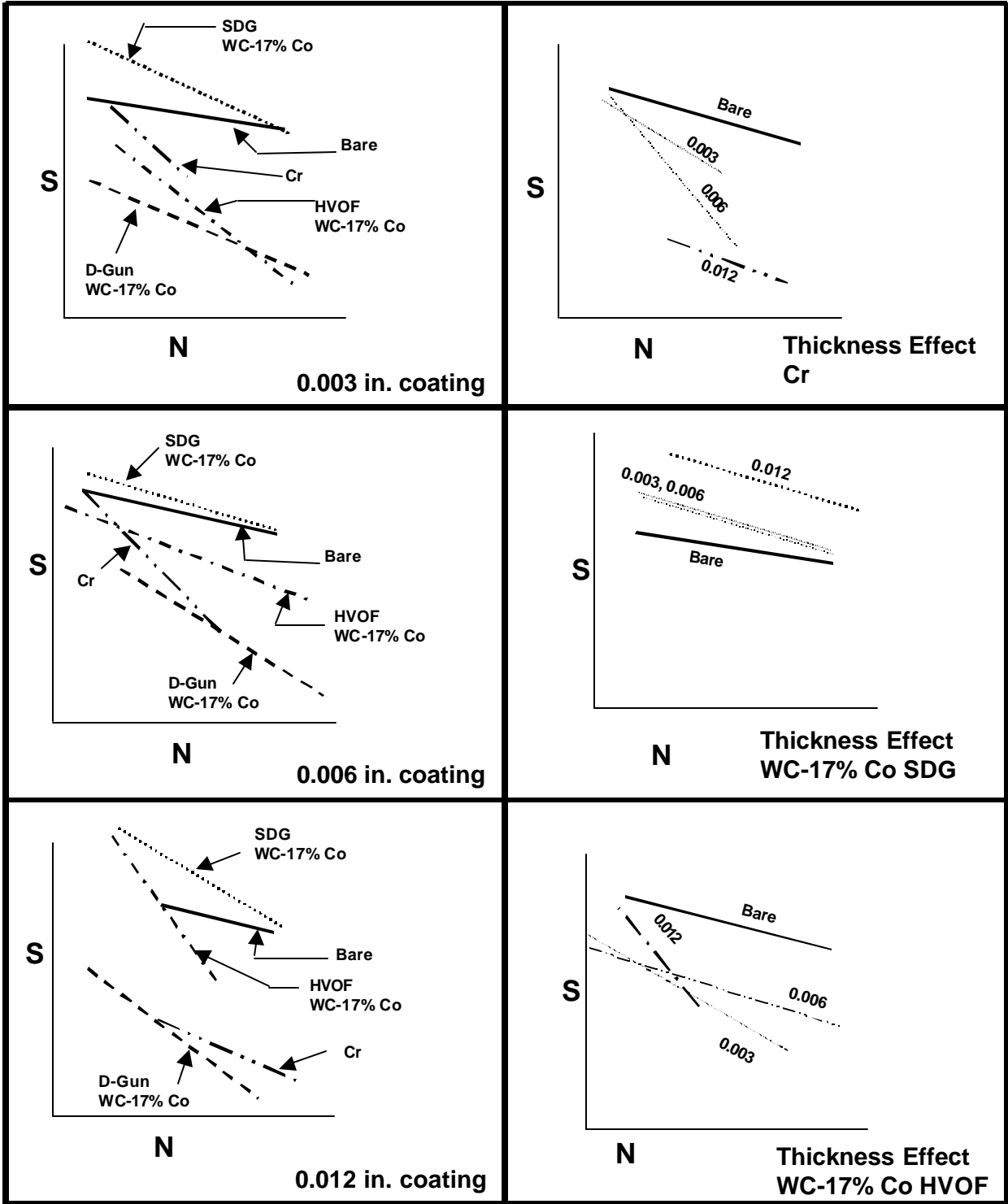
Figure 4: LBD Axial Fatigue Data. Hy-Tuf (F$_{tu}$: 220-250 ksi). Smooth Bars. Schematic

was not sufficiently adherent (i.e., improperly applied). Another observation here is the 0.003-in. plating fatigue curve, which intersects with the 0.012 and bare metal curves. This would mean that at high stresses, the 0.003 plating involves more debit than the 0.012 plating. At low stresses, by contrast, the 0.003 plating would involve an inexplicable fatigue credit. Evidently, the fatigue data for Cr plated 300M are questionable.

(d) Based on (a) and (c), it can be concluded that as thickness is increased, performance of HVOF coatings will improve, while that of Cr plating will deteriorate. While these trends make sense, it must be emphasized that the true ranking of coatings with respect to each other and to the bare steel depends critically on the bond strength achieved. A coating that ordinarily involves a fatigue debit may show no debit at all if it is nonadherent. As the achieved bond strength increases, the debit will tend to increase. Similar arguments can be developed for coatings that ordinarily involve a fatigue credit.

(e) The data indicate that for thinner coatings, especially in the low cycle (high stress) regime, Cr plating can involve somewhat lesser debits than those of HVOF coatings. Based on the arguments in (d), it is felt that this trend is brought about, at least in part, by Cr adhesion/bond strength problems. This anomalous trend is reversed for thicker coatings and/or in the high cycle (low stress) regime.

(f) The curves of Figures 3 and 4 tend to intersect. This is true for curves of the same coating at various thicknesses and for different coatings of the same thickness. These intersections could be, in part, a result of data scatter and/or curve fitting problems. In some cases, however, it is likely that they are real, reflecting actual transitions in fatigue behavior. It is thought that these transitions are brought about by changes in the magnitude of the fatigue debit/credit, as the applied stress is changed. The debit/credit changes, in turn, could be due to changes in the property that dominates fatigue behavior; viz., strength dominates the low cycle regime while ductility dominates the high cycle regime. A possible culprit here could be changes in crack initiation sites [coating, interface, near subsurface or bulk].

B2.1.7.3 Follow-Up on Tensile & Fatigue Results

(a) HVOF: The fatigue performance of the HVOF coating in the LBD effort was deemed by many as disappointing; the coating was applied at Southwest Aeroservice, Inc. (Tulsa, OK). In a follow-up letter, [6] Southwest explains that this disappointing performance was because the work was performed early on in the technical life of HVOF, before all processing variables were optimized. A culprit in the poor showing, the letter points out, was the circumferential polishing used in specimen preparation. The letter goes on to indicate that in more recent (May 1998) work, for Boeing-Seattle, Southwest used coarser WC-Co (presumably, WC-17%Co) powders. The resulting fatigue performance of the HVOF coating was superior to both Cr plating and the bare materials, enabling Boeing-Seattle to qualify Southwest for current production; in later communication, the author indicated that the material investigated was AISI 4340M (300M) steel. The letter adds that, in 1999, Southwest performed independent research to investigate the performance of a new WC-Co-Cr (possibly, WC-10%Co-4%Cr) coating on shot peened AISI 4340 steel. The WC-Co-Cr coating was applied using two HVOF systems and optimized process parameters. The HVOF coating, as applied by the two systems, performed similarly not only to each other but also to the bare base metal and markedly better than Cr-plate. To corroborate, the letter refers to an attached graph. The graph proved difficult to interpret, due to overlapping data points and what appears to be an inconsistent legend. The data, however, seem to indicate that, at lower stress levels Cr plating is superior to all HVOF coatings.

Based on the above and the results of other efforts, LBD is contemplating a "delta" test program to once and for all confirm the merits of HVOF. The program will focus on tensile and fatigue testing of bare and HVOF coated 300M and Hy-Tuf steels. The coating of choice is WC-10% Co-4% Cr, and it will be applied at Southwest Aeroservice, Inc. The coating thicknesses to be investigated are 0.003 and 0.012 in. for the tensile specimens and 0.003, 0.006 and 0.012 in. for the fatigue specimens. All specimens will be shot peened, prior to coating.

(b) D-Gun: Fatigue performance of D-Gun coatings was the worst among all deposits, including Cr plating. By contrast, fatigue performance of SDG coatings was the best among all coatings. This was rather unexpected, since both types of coatings were applied at the same facility. Mr. John Quets (317/240-2633) of Praxair was contacted to explore the issue with him. He indicated that the reason for this poor showing is that the coating selected by LBD (WC-17% Co) is not ideal for D-Gun applications involving fatigue; other D-Gun coatings would have fared much better. As to how the same coating, applied by SDG, displayed the best fatigue performance, he indicated that SDG is the new technology, while D-Gun is old technology. Since D-Gun and SDG are not contenders in the Cr replacement effort, the issue was not further explored. D-Gun users, however, may wish to carefully assess the LBD results.

B2.2 Phase 2, In-Flight Testing

This phase has not yet commenced and no information is available.

B3.0 THE BOEING AND AIRBUS INDUSTRIES EFFORTS

There are significant ongoing Cr replacement efforts at both Boeing and Airbus Industries. It was not possible, however, to locate publications, in the open US literature, detailing these efforts; segments of the Boeing effort, however, appear to be linked to and are referenced in the DOD-HCAT effort. [1,2] A very brief account of the efforts can be found in reference 6. It appears that both efforts have been initiated by requests from Lufthansa. The Boeing effort started in 1995 and that of Airbus Industries, through Messier-Dowty, in 1996. Both efforts, as in the DOD-HCAT case discussed earlier, involve two phases. Phase 1 is laboratory testing and Phase 2 is in-flight testing, and the initial focus is on landing gear applications.

B3.1 The Boeing Effort

Reference 7 presents a brief account of the Boeing effort, without presenting actual data. Basically, this reference indicates that an extensive thermal spray coating test program had been underway at Boeing for several years before the Lufthansa request. This program involved different substrates (steel, aluminum and titanium) and different application processes. The testing involved fatigue, bond strength and wear (sliding, erosion, abrasion and impact). The satisfactory performance of HVOF WC thermal spray coatings allowed Boeing to begin Phase 2 in early 1996. Phase 2 of the decision making process involved in-flight testing of HVOF coated 737 nose landing gear inner cylinder. After multiple inspections, the general experience is generally positive.

Additional information on the Boeing effort was gleaned from discussions with a member of LBD Materials Engineering, familiar with that effort, and from a cursory review of test protocols as well as some preliminary data and airline inspection reports. A brief account is presented below.

It appears that, in addition to what is listed in reference 7, Phase 1 included extensive metallographic examination and corrosion testing. Bond strength determinations were performed using a setup that is similar to ASTM C633. Smooth and notched bar fatigue data were generated for Cr plating and WC-17% Co and WC-10% Co-4% Cr coatings, applied by HVOF, D-Gun and SDG. The fatigue specimens were made of heat treated and shot peened AISI 4340. In general, Cr plating appears to be associated with a fatigue debit. The data suggest that HVOF and SDG coatings could involve either a fatigue debit or credit. The ranking of the coatings with respect to each other and to the bare steel seem to be influenced by the stress level and by whether the specimens are notched or smooth. The data seem to indicate that specimens coated by different processors, using the same generic process (HVOF), can have different fatigue lives, in spite of the fact that the powders were procured to the same specification.

In-flight testing (Phase 2) is underway, using landing gear components, coated with WC-17% Co or WC-10% Co-4% Cr. In-service inspections are performed by Lufthansa, Delta and British Airways. In general, the experience with the thermally sprayed coatings seems to be a positive one. It appears, however, that there was some wear issue with the WC-10% Co-4% Cr coating. The nature of this issue, however, could not be discerned.

The Boeing effort seems to have generated an impressive collection of data over the years. It is most unfortunate that such data have not been published in the open literature.

B3.2 The Airbus Industries Effort

Reference 7 describes, in some detail, the Airbus Industries effort. A brief account is presented below.

Phase 1, undertaken by Messier-Dowty included comparisons between WC coatings and Cr plating in five areas; viz., corrosion, machinability and detectability of grinding defects, friction and wear, fatigue and performance against seals. The corrosion and fatigue tests, respectively, were performed by Lufthansa Technik and Messier-Dowty, and Southwest Aeroservice prepared the specimens for both tests; one of the authors of the reference is listed as an employee of Southwest. In addition to Cr plating, WC-17% Co and WC-10% Co-4% Cr coatings, applied by HVOF or SDG, were investigated. The plating/coating thicknesses investigated were 0.004 and 0.008 in.; the thermally sprayed coatings, however, were deposited oversize and then ground.

Corrosion testing was performed per ASTM B117 (salt spray), using AISI 1010 steel specimens; thicker specimens were used for the thermally sprayed coatings. The graph, summarizing the corrosion results, presented in the reference contains some inconsistencies, making a full interpretation difficult. It appears, however, that, for both coating thicknesses, the HVOF WC-10% Co-4% Cr coating outperformed the HVOF WC-17% Co coating which, in turn, outperformed Cr plating; performances of the SDG and HVOF WC-10% Co-4% Cr coatings appear to be similar. Cr plating, on the other hand, outperformed the SDG WC-17% Co coating.

Fatigue testing was performed on HVOF coated and Cr plated specimens, made of E35NCD16, a French ultrahigh strength steel; the applicable strength range is not listed. The plated and coated specimens were shot peened prior to coating; it is not clear, however, whether or not the bare specimens were shot peened. The results of the fatigue testing are depicted schematically in Figure 5; the applicable coating thickness investigated was not listed. The data indicated that the fatigue performance of the HVOF coating and the Cr plating are more-or-less similar; in fact, the data indicate that Cr plating involves somewhat lesser debit than the HVOF coating.

Wear testing revealed that HVOF WC coatings were at least as good as Cr plating. The nature of these wear tests (abrasive or sliding), however, was not indicated.

Based on the results of Phase 1 testing, Messier-Dowty allowed the in-flight evaluation (Phase 2) to proceed. Phase 2 involves a sliding member of the A320 nose landing gear. No other details of that phase were disclosed in the reference.
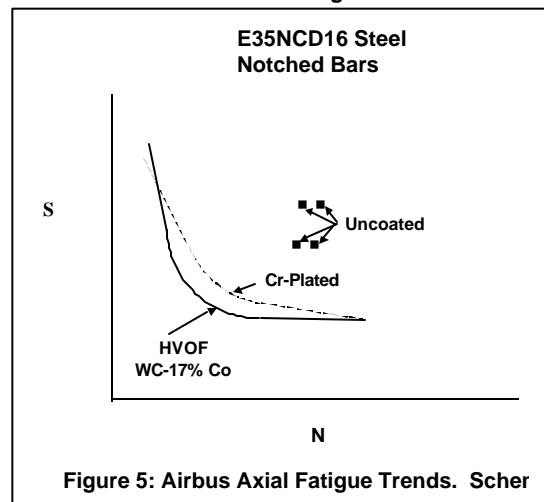

Figure 5: Airbus Axial Fatigue Trends. Scher

B4.0 ANALYSIS OF THE FATIGUE DATA

The Boeing and LBD data indicate that SDG coatings can give rise to a fatigue credit or debit. The Boeing data suggest a similar possibility for HVOF coatings. It is generally believed that a fatigue credit would be due to a peening effect, resulting from the carbide particles, in the gas stream, impacting the metal surface and inducing residual compressive stresses; for simplicity, will refer to this as powder peening. Implied here is that the carbide particles remain in the solid state (i.e., do not melt), so as to impart a peening action; only the Co or Co+Cr matrix would melt. As indicated in reference 6, coarser carbide powders enhance the peening action. Powder peening is possible in the HVOF and the SDG/DG processes, due to the relatively low application temperatures and the short dwell times at those temperatures. In almost all other spray processes, no peening takes place since melting of the carbide particles is inevitable due to the fairly high application temperatures and the relatively long dwell times that are typically involved.

The HVOF heat input, during deposition, also affects fatigue performance. The LBD tensile data indicate that this heat input can reduce (degrade) substrate strength below acceptance levels; the implication here is that the heat input had affected the entire cross section, not just a surface layer. The degraded strength values are typical of tempering at temperatures a few hundred degrees above the

recommended ranges (i.e., over-tempering). A reduction in tensile strength would tend to promote a fatigue debit, especially in the high cycle (low stress) regime. Apart from degrading substrate properties, the HVOF heat input has a direct effect on residual stress fields. Cooling and solidification of the partially molten HVOF deposit would generate adverse residual tensile stresses. While these tend to be localized in the coating, substrate layers beneath the coating could be affected. In these layers, therefore, the residual tensile stresses, resulting from cooling, will be superimposed on the residual compressive stresses, resulting from powder peening.

There is more to the HVOF heat input. In addition to degrading substrate strength, the over- tempering process can lead to two additional phenomena. The first phenomenon is an increase in tensile elongation (i.e., an increase in ductility). This would tend to improve performance and promote lesser fatigue debits, even credits, in the low cycle regime. The improvement, however, would be insignificant, since the increase in ductility, associated with the slight over-tempering involved, would be very small. The second phenomenon is the possibility of relieving existing and developing residual tensile/compressive stresses, as a result of the HVOF heat input. The higher the substrate temperature attained, and the longer the time at that temperature, the more complete this stress relief would tend to be. It is thought, however, that the contribution of this stress relief would be a minor one, due to the typically short coating times and the fairly low temperatures attained.

A note on the initial residual stress state of the specimens, prior to coating, is in order here. In specimens that are not shot peened after heat treatment, the surface layers of the steel would typically be under residual tensile stresses. These will tend to augment the adverse cooling stresses. In shot peened specimens, the surface layers of the steel would be under residual compressive stresses. These will generally augment powder-peening stresses. It is to be noted, however, that fatigue debits and credits of coated specimens are measured relative to bare specimens having the same initial state. As such, the initial residual stress state of the specimens should not be a deciding factor as to whether a credit or a debit would result from coating.

One detail that should not be overlooked is the load carrying capacity of the deposit, be that Cr plating or some HVOV coating. So long as the deposit remained intact and adherent to the substrate, it will carry a portion of the acting load. For axial elastic loading and round specimens, the load fraction, f, carried by the coating is $f = 1 / \{[E_S/ E_d] \, [r^2/t \, (2r+t)] + 1\}$. In this formula, r is the radius of the specimen, t is the deposit thickness and E_S and E_d, respectively, are the elastic (Young's) moduli for the substrate (specimen) and deposit. It can be shown that increasing the deposit thickness and/or its elastic modulus, relative to that of the substrate, would increase the load fraction carried by that deposit. To illustrate, consider a round, 0.25 in-diameter steel specimen (E_S = 30 MSI), coated with some HVOF WC-Co coating (E_d = 21 MSI) that is 0.003 in. thick. The load fraction carried by this 0.003-in. deposit would be about 3%. This fraction would increase to about 13% for a 0.012-in. thick coating. The corresponding fractions for Cr plating (E_d = 40 MSI), applied to an identical specimen, would be 6% and 21%, respectively, for the 0.003 and 0.012 in. thick deposits. In all cases, the load fraction, carried by the deposit, becomes less significant as specimen size (r) is increased; e.g., considering a 0.012-in. thick Cr plating, an increase in specimen diameter from 0.25 to 1.00 in. reduces f from 21% to just about 6%.

Based on the above analyses, it is possible to examine the effects of increasing coating thickness on the fatigue debit/credit associated with HVOF coatings. There are competing factors to be considered here. On the one hand, the load carrying capacity of the coating and powder peening effects tend to promote credit. On the other hand, the HVOF heat input tends to promote debit. We now consider each factor in some detail. Increasing coating thickness is accompanied by an increase in its load carrying capacity. This would translate to less stress in the steel substrate which, in turn, would be manifested in the form of a progressively increasing fatigue credit, as the coating thickness is increased; the credit would become less significant as specimen size is increased. It can also be argued that increasing the coating thickness would somewhat enhance the powder peening action, due to increased coverage and, possibly, intensity. This enhancement would also be manifested in the form of a fatigue credit. Fast deposition rates, especially when thicker coatings are required, involve large heat inputs. It can be argued, therefore, that, under such circumstances, increasing the coating thickness would lead to progressive degradation in substrate strength. Indeed, this is what the LBD tensile data show. A progressive degradation in strength would, in turn, tend to promote a progressively increasing debit as coating thickness is increased, especially in the low stress regime. Increasing heat inputs is also expected to increase the magnitudes of the adverse tensile stresses, resulting from cooling of the partially molten coating. This should also promote a progressively increasing debit as coating thickness is increased.

Thus, it appears that, for HVOF coatings, whether a fatigue debit or credit would be observed and the magnitude thereof will depend on three factors. The first is the residual compressive stresses, resulting from powder peening. The second is the load carrying attributes of the coating; this factor becomes less significant as specimen size is increased. Both of these would favor fatigue credit. The third is the HVOF heat input, which would favor fatigue debit, due to the possible degradation of substrate strength and the introduction of adverse cooling residual stresses. The magnitude of the fatigue debit/credit observed is the algebraic sum of the credits and debits resulting from these three factors. As such, depending on processing details, there should be debit-type and credit-type HVOF coatings. In debit-type coatings, increasing the coating thickness should be accompanied by a progressively decreasing debit, which could eventually turn into a progressively increasing credit; this appears to be the case in the LBD effort. In credit-type coatings, increasing the coating thickness should be accompanied by a progressively increasing credit.

The situation for Cr plating is simpler. There are no temperature or powder peening effects to contend with. As such, there are only two competing factors. The first is the residual tensile stresses associated with Cr plating, which would promote fatigue debit. The second is the load carrying capacity of the plating, which would promote fatigue credit. As coating thickness is increased, the first would favor a progressively increasing debit, while the second would favor a progressively increasing credit. Data cited herein, however, consistently depict fatigue debits for Cr plating. Furthermore, LBD data indicate progressively increasing debits, as plating thickness is increased. It must be, therefore, that the residual tensile stress factor far outweighs the load carrying attributes of the plating. One reason for this may be that the load carrying attributes are undermined by the presence of micro/macro cracks.

The data shown in Figure 5 and some of the data in Figures 3 and 4 depict Cr plating as having a lesser fatigue debit than the HVOF WC coating. It is felt that this is an indication of plating and/or HVOF processing problems. For example, if insufficiently adherent Cr plating is deposited, the resulting residual stresses will not develop to their full extent. Consequently, the apparent fatigue debit would be less than what it should be. In the extreme case of nonadherent Cr plating, no debit at all would be observed. Similarly, an HVOF coating would display a larger debit, or a lesser credit, than normal if the heat input during coating degrades substrate strength and/or introduces high magnitudes of adverse cooling stresses. The opposite effects might be observed, if the coating were insufficiently adherent.

A comment on the NAWC data is in order here. In conventional fatigue testing, both the Cr and the HVOF coatings displayed inferior performances compared to the bare steel; conventional means testing in air, as in all the other efforts. In corrosion fatigue testing, however, the bare steel displayed inferior performance compared to both coatings. It must be, therefore, that the corrosion protection afforded by either coating has prevailed over all the other factors that normally affect conventional fatigue performance. Note that all specimens were coated to the same thickness. Thus, for either Cr or HVOF, the load carrying attributes of the coating would be the same in both fatigue and corrosion fatigue specimens.

Appendix B References:

(1) Bruce D. Sartwell and Philip E. Bertz, Advanced Materials & Processes, ASM, August 1999, pp 25-28.
(2) Bruce D. Sartwell, Welding Journal, July 2000, pp 39-43.
(3) Jerry D. Schell and Mark Rechsteiner, Plating & Surface Finishing, July 2000, pp 17-23.
(4) Based on a presentation by E.U.Lee (301/342-8069) of NAWC, at the AeroMat 2001 conference, which was held June 11-14, 2001 in Long Beach, CA.
(5) Mary Ann Forrest-Woodward, Charles T. Bishop and Thomas P. Posten, The Chromium Plating Replacement Program at Boeing-Long Beach Division, Published by the American Electroplaters and Surface Finishers (AESF) Society, 1999. Presented at the 1999 Aerospace/Airline Plating and Metal Finishing Forum.
(6) Letter from Jim Nuse (Southwest) to John A. Falk (Boeing-Seattle), dated 18 October 1999.
(7) B. E. Bodger, R.T.R. McGrann and D. A. Sommerville, Plating & Surface Finishing, September 1997, pp 28-31.

DISTRIBUTION

Gregory S. DiLebero   FAA / ANM-120L

Thomas P. Posten   Boeing, Long Beach, CA

Bruce D. Sartwell   Naval Research Laboratory, Washington, DC

Eun Lee   Naval Air Warfare Center, Patuxent River, MD

Jack Cummins   Delta Airlines, Atlanta, GA

Jim Nuse   Southwest Aeroservices, Inc., Tulsa, OK

Walter Mroch   US Chrome, Long Beach, CA

David Swartz   FAA / ANM-108S

Brian Keeley   Keeley Associates, San Juan Capistrano, CA

Tom Yentzer   Robins AFB, GA

Christian Beaufils   Airbus Industries, Toulouse, France