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Solid particle erosion mechanisms of hard protective coatings

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ABSTRACT

In order to assess the material loss mechanisms of monolithic coatings subjected to solid particle erosion (SPE) using angular alumina particles with relatively low velocities (<100 m/s), we studied the erosion behavior of several hard coatings deposited by pulsed DC magnetron sputtering. We first validated a new methodology for the measurement of volume loss and optimized the testing conditions to obtain a measured erosion rate (*ER*) free from experimental artifacts. We then correlated the measured *ER*s to the mechanical properties, measured by depth sensing indentation, and found that the *ER* was strongly dependent on the target hardness (H_t) of the materials ($ER \propto H_t^{-6.8 \pm 0.5}$). In order to understand the material loss mechanisms, we studied three coating systems in greater detail with the help of fracture characterization and a morphological study of the eroded surfaces. It was found that fracture toughness was not a good predictor of *ER* and that the material removal was the result of ductile indentation and cutting. Finally, in an effort to understand the role of particle fracture, we measured the particle size distributions of the powders before and after erosion testing and found that particle breakup was proportional to the target hardness but not sufficiently large to explain the large drop in *ER*s with increasing H_t.

1. Introduction

When aircraft operate in harsh environments where hard particulate matter is entrained by the air flow into the operating engine, severe wear of exposed components may occur through material removal by solid particle erosion (SPE). This type of damage is most prominent in the first stage of the aircraft engine, where the compressor blades can be eroded to such an extent that aerodynamic performance and even structural integrity are compromised. Consequently, much work has been done in academia and industry in order to understand the material loss mechanisms present in SPE and to develop protective approaches that will increase component lifetimes. One such technology is the use of hard protective coatings to impede the erosion of the predominantly metallic engine components.

Of the many different coating systems proposed for protection against SPE, two main categories emerge: TiN-based and carbon-based. Carbon-based films are predominantly thick diamond coatings deposited by chemical vapor deposition [1–3]. Due to their very high hardness, these have been found to be extremely resistant to SPE impact damage [3,4]. However, they are difficult to implement in practice since they need to be deposited at temperatures that exceed the permissible limits of the metallic components. For that reason, the most frequently used coatings for SPE protection are TiN-based in monolithic [5–9] or multilayer [10–12] forms, which can be deposited on technologically relevant substrates.

Since the early 1990s, Ti/TiN multilayer systems have been used because their multilayer design offers the possibility of depositing thicker coatings by relaxing residual stresses and of enhancing erosion resistance. In fact, it has been shown by Borawski et al. [11] that multilayer architectures are more resistant to SPE when the stress field generated during impact encompasses a larger volume of the coating (large rounded particles) because of the crack tip blunting effect of the ductile layers and the deflection of cracks at interfaces. On the other hand, monolithic coatings are more durable under SPE by small, hard and angular particles because the damage is generally confined to a small volume in the top layer and therefore, the ductile interlayers of the multilayer coating would not be beneficial to erosion resistance.

The SPE resistance of monolithic coatings can be enhanced by using materials with high toughness to prevent crack growth, and high hardness to inhibit crack initiation by dissipating the particle kinetic energy through fragmentation and minimizing the penetration of the particle on impact. Consequently, alloyed and nanocomposite TiN-based systems are being investigated in greater numbers. One structurally hardened system that is now being implemented industrially is TiAlN, which has been shown to not only possess high SPE resistance [8,13], but also good stability at higher temperature [14–16]. However, more recently, nanocomposite coatings have been shown to be promising candidates for SPE resistance because of their very high hardness and ability to slow crack propagation [17–19]. In fact, CrSiN [20], TiSiN [21] and TiSiCN [22,23] systems have been shown to be highly resistant to SPE by small angular alumina particles.

In order to select the appropriate coating system, the material loss mechanisms must be properly understood. In the first place, the metallic surfaces to be protected (stainless steel or titanium alloys) erode in a

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predominantly plastic mode of material removal through micro-cutting or plowing mechanisms [24–26]. On the contrary, the hard ceramic protective coatings are brittle in nature and will present much more complex erosion behaviors. In fact, it has been shown that brittle materials exhibit surface removal mechanisms very similar to those encountered during indentation [27]. In the case of large round particles, Hertzian cracks will form and can coalesce into deep wormhole-like craters [28,29], while for sharp particles, the damage can progress from plastic cutting or plowing of the surface for low energies to radial/median cracking at higher energies and finally to lateral fracture removing significant amounts of matter [30–32]. The transitions between mechanisms are a function of the coatings' corresponding elastic (Young's modulus [E]), plastic (hardness [H]) and fracture (fracture toughness [K_{Ic}]) properties [27].

In addition, it has been shown that the erodent particles can be crushed or fragmented upon impact when the hardness of the target (H_t) is greater than that of the particle (H_p) [27,33–35]. As a result, the particle breakup may lead to very low erosion rates and may also cause a change in the material loss mechanism. For low H_p/H_t ratios, and low particle velocities, a micro-chipping mechanism has been observed by several authors [11,27,33] and has generally been characterized by a smooth surface after erosion when compared to the very rough surfaces resulting from lateral fracture of brittle materials [36].

While it has been observed that coatings exhibit the different brittle fracture modes (Hertzian, radial and lateral cracking) when impacted by large and/or highly energetic particles, there is no description of the mode of material loss when eroded by small angular particles with relatively low velocities (<100 m/s).

In the present paper, we study the above-mentioned mechanisms for several hard coatings deposited by pulsed DC magnetron sputtering. We first validate a new methodology for the accurate measurement of volume loss and we then correlate the measured erosion rates to the material parameters measured by depth-sensing indentation. Furthermore, in order to understand the material loss process, we study three of the coating systems in greater detail with the help of fracture characterization and a morphological evaluation of the eroded surfaces. Finally, in an effort to assess the role of particle fracture, we determine the particle size distributions of the powders before and after erosion testing.

2. Experimental methodology

2.1. Coating deposition

All coatings were deposited by reactive pulsed DC magnetron sputtering (PDCMS) in a vacuum system equipped with two magnetrons. The base pressure of the reactor, described in detail in [37], was below 1×10^{-6} Torr. Three different substrates were used: single crystal silicon (c-Si), Ti–6Al–4V (Ti64) and AISI 410 stainless steel (SS410). The metallic substrates were polished to a mirror-like finish and all substrates were cleaned in acetone and isopropanol using an ultrasonic bath. Then, the substrate surfaces were sputter cleaned in a capacitively coupled radio-frequency (RF) Ar plasma at a negative bias (V_b) of −600 V, in order to remove the surface oxide layer prior to deposition. Depending on the type of coating, the substrate deposition temperatures vary between 250 and 400 °C while V_b was between −50 and −200 V (summarized in Table 1).

In all cases, a thin chromium or titanium adhesion layer (<0.5 μm) was first deposited. This was followed by the principal monolithic coating. Here we study binary (CrN and TiN), ternary (CrSiN and TiSiN) and quaternary (TiSiCN) material systems. The silicon content in the ternary coatings was adjusted by changing the current on the Si target and the carbon content in the quaternary coating was controlled by the flow of CH_4 during deposition. The thicknesses of the deposited films (t) were between 6 and 13 μm, as determined by scanning electron microscopy (SEM, JEOL JSM7600F). The coating deposition parameters and thicknesses are presented in Table 1.

Table 1
PDCMS deposition parameters and thicknesses of the coatings.

Coating	Substrate	Deposition temperature (°C)	Bias, V_b (−V)	Thickness, t (μm)
CrN-1	c-Si	250	100	8
CrN-1	Ti64	250	100	8
CrN-2	SS410	300	200	13
CrSiN-1	SS410	300	200	10
CrSiN-2	SS410	300	200	8
CrSiN-3	SS410	300	200	9
TiN-1	SS410	400	200	10
TiN-2	c-Si	400	100	8
TiN-2	Ti64	400	100	8
TiSiN-1	c-Si	400	50	6
TiSiN-1	Ti64	400	50	6
TiSiN-1	SS410	400	50	6
TiSiN-2	SS410	400	200	10
TiSiCN-1	SS410	400	200	10

2.2. Elasto-plastic properties

The elasto-plastic properties of the target materials were evaluated by depth-sensing indentation (Hysitron Triboindenter). The coating hardness (H_t) and reduced Young's modulus (E_r) were measured using a Berkovich geometry indenter and the data was analyzed using the widely accepted Oliver and Pharr [38] methodology. The tip geometry and system compliance were calibrated using a fused silica standard, and the system drift, typically less than 0.1 nm/s, was measured before each indentation by maintaining the tip in contact with the surface and monitoring the drift for 40 s. Each coating was probed by a 3 × 3 indentation array with a spacing of 100 μm between indentations. Each indentation was composed of a 25 cycle partial-unload load-function with the maximum load of each cycle (P_{max}) increasing from 10 mN to 1000 mN. For each load cycle, the load increased to P_{max} over 5 s followed by a 2 s holding segment, and finished with a 5 s unloading segment down to 80% of P_{max} (Fig. 1). It should be noted that this type of indentation load-function was found to be valid for the hard ceramic coatings studied because they do not present strain hardening under the loading conditions considered here.

After removing failed indentations (generally caused by tip slippage), we used all the remaining valid partial-unload indentations for each coating to calculate the average elasto-plastic properties as a function of indentation contact depth (h_c) as well as the standard deviation. For each curve, the Young's modulus was calculated by extrapolating the linear part of the E_r curve (approximately h_c/t <0.1) to $h_c/t = 0$ and the hardness by averaging the plateau portion of each H_t curve. In this manner, we obtained a depth profile of the elasto-plastic properties for each indentation.

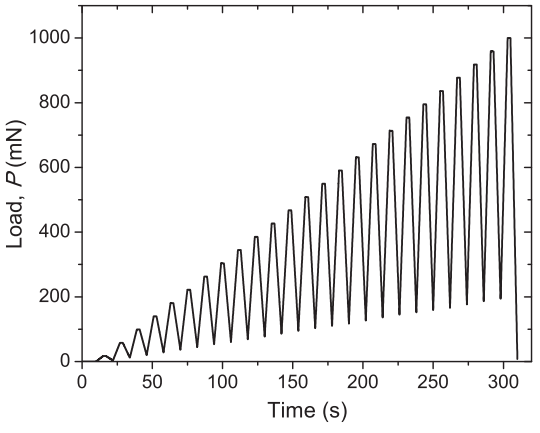

Fig. 1. Partial-unload load-function used to obtain the coating mechanical properties.