



High-Temperature Solid Particle Erosion of Aerospace Components: Its Mitigation Using Advanced Nanostructured Coating Technologies

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Abstract: Solid particle erosion of gas turbine blades in the aerospace sector results in increased maintenance costs, high pollution, reduced engine efficiency, etc. Gas turbines in aircraft are usually operated at high temperatures. Based on the compressor stage, the temperature varies from 100–600°C, whereas turbine blades, after combustion, experience a very high temperature between 1000–1400 °C. So, a better understanding of temperature-dependent solid particle erosion is required to develop suitable solid particle erosion-resistant coatings for gas turbine blades. In this review, a detailed overview of the effect of temperature on the solid particle erosion process and different types of erosion-resistant coatings developed over the last four decades for compressor blades are discussed in detail. In the initial sections of the paper, solid particle erosion mechanisms, erosion by different erodent media, and the influence of erosion on gas turbine engines are discussed. Then, the erosion rate trend with increasing temperature for ductile and brittle materials, high-temperature erosion tests in a corrosive environment, and the role of oxidation and bonding nature in high-temperature erosion are examined. In most cases, the erosion rate of materials decreased with increasing temperature. After this, the evolution of erosion-resistant coatings over the last four decades that are first-generation (single-phase coatings), second-generation (metal/ceramic multilayer coatings), and third-generation (nanocomposite and nano-multilayer coatings) erosion-resistant coatings are reviewed in detail. The third-generation nano coatings were found to be superior to the first- and second-generation erosionresistant coatings. Finally, some of the commercial or notable erosion-resistant coatings developed in the last decade are discussed. The paper concluded with the research gaps that need to be addressed to develop efficient erosion-resistant coatings.

Keywords: solid particle erosion; temperature dependent erosion; gas turbine blades; erosion resistant coatings; PVD coatings for erosion resistance

1. Introduction

1.1. Solid Particle Erosion

It is known that solid particle erosion (SPE) is a serious problem faced by gas turbine engines, wind turbines, turbines in hydroelectric power plants, etc. The gas turbine engine is one of the most important and critical parts of the aircraft (Figure 1) [1,2]. Solid particle erosion in gas turbines results in increased maintenance costs, unscheduled removals, and high fuel consumption/emissions. Apart from this, mission completion rates, safety, and readiness decrease. In extreme cases, such as during long flights in a harsh environment, failure of the engine and catastrophic accidents can also happen [3–9]. In the case of wind turbines, apart from rain erosion, SPE is also a serious problem. The wind blades are usually made of fiber-reinforced polymer composites, which are the most expensive part of wind turbines [10]. The erosion of wind turbines results in high costs and a reduction in annual energy production. Erosion-resistant protective coatings, such as polyurethane-based



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Copyright: © 2022 by the authors. Licensee MDPI, Basel, Switzerland. This article is an open access article distributed under the terms and conditions of the Creative Commons Attribution (CC BY) license (https:// creativecommons.org/licenses/by/ 4.0/). coatings, are commonly used on turbine blades due to their mechanical strength, elasticity, long-term damping, thermal stability, excellent ultraviolet (UV) stability, chemical resistivity, and low toxicity [10–13]. Similarly, hydroelectric power is another significant renewable energy source [14,15]. Here, the SPE and corrosion of the turbine are also challenging problems. Solid particles in rivers, such as quartz and feldspar from the Asian mountain range, cause the hydro turbines to suffer from erosion. Consequently, maintenance costs and production losses increased considerably [16,17]. Erosion and corrosion-resistant coatings were found to be one of the solutions to this problem. The coating improves the turbine's performance and its lifetime [18,19]. However, in the present review, we would like to focus on the SPE damage of gas turbines used in aerospace, temperature dependent SPE, and the protective coatings developed for compressor blades over the last four decades.



Figure 1. A modern jet engine used to power Boeing 777 aircraft. Reprinted with permission from ref. [1]. Copyright 2004, Elsevier.

Solid particle erosion involves different wear mechanisms that occur during impact between relatively small solid particles and mechanical components [3,4]. The erosion mechanism depends on erodent particle properties (viz., hardness, size, shape, direction, and velocity) and the properties of eroded material (ductile and brittle). The erosion rate increases with increasing erodent velocity, and these two parameters are related according to Er α Vn, where Er is the erosion rate and V is the particle velocity [20]. Researchers in different studies have reported values of n ranging from 2.0 to 3.0, depending upon the size of erodent particles [20]. Finnie et al. [21] studied the erosion properties of different metals with silicon carbide particles at an impinging angle of 20° and concluded that the velocity exponent (n) is between 2.05 and 2.44. Apart from the velocity, an increase in erodent particle size results in increased erosive damage for ductile materials until a saturation level is reached, after which no further change is indicated [22]. The shape of the erodent particles also plays an important role in erosion damage, where coarse particles lead to greater erosion than smooth, spherical particles [23]. Overall, erosion type and rate vary depending on the erodent flow conditions. Typical erodent particles of 50 µm angular alumina, silica sand, and coarse Arizona road dust are shown in Figure 2 [24]. Among these, 50 µm angular alumina is widely used in erosion tests, as suggested in the ASTM-G76 standards.



Figure 2. Typical erodent particles (**A**) Angular alumina, (**B**) silica sand, and (**C**) coarse Arizona road dust. Reprinted with permission from ref. [24]. Copyright 2009, Elsevier.

In addition to the erodent flow conditions, erosion damage also greatly depends on the material surface when struck by particles. Brittle and ductile material erodes differently for the same flow conditions. Figure 3 shows the erosion rate with increasing impinging angles for ductile and brittle materials [25]. It was observed that for high impact angles, the soft and ductile materials are more erosion resistant than the hard and brittle materials, and vice versa at low impact angles [20,21,24,25]. This is because the failure mechanisms are distinctly separate for these two kinds of materials (Figure 4) [26]. In the case of ductile materials, high erosive wear occurs due to plastic cutting and shear deformation (Figure 4). Due to this, at lower impact angles, ductile materials lose more material [5]. Whereas failure in brittle materials occurs through fatigue and crack network formation (Hertzian fracture). So, these materials lose more weight during normal impact (Figure 4).









1.2. Gas Turbine Engines and Solid Particle Erosion

The working principle of a gas turbine is the Brayton cycle, with air as the working fluid. The compressor in the gas turbine sucks in the atmospheric air and increases its pressure. Following that, heat is added to the compressed air by spraying fuel and igniting it. As a result, combustion generates a high-temperature pressurized gas flow, and it enters the turbine. Some of the generated energy is used to drive the compressor, and the rest of the energy comes out in the exhaust gases, which can be utilized for external work. Gas turbine engines significantly contributed to global economic growth [27,28]. However, gas turbines face unavoidable serious challenges from SPE, fouling, and corrosion while operating in harsh environments such as deserts, forests, and near volcanoes [29–34]. During operation in these environments, erodent particles such as sand, dust, coal derivatives, and fly ashes enter the gas turbine and affect the stator vanes, rotor, and turbine blades [29–37]. Initial studies on erosion originated during the Second World War period [38]. However, studies to understand the concept of erosion or the mechanism of material removal started from 1960 onwards [3–6,39]. Theoretical analyses were done to understand and correlate the erosion of different materials with varying parameters. Many theoretical models were developed for various kinds of materials to predict erosion behavior [3–7,39]. As the compressor is the starting point for erodent particles, the compressor blades undergo geometric modifications, and pressure losses occur [8,20,29–32]. Figure 5 shows the damaged compressor blades due to SPE [32]. Damage to the compressor blades by erodent particles usually occurs by removing the material, whereas turbine blades face fouling. High temperatures at the turbine cause erodent dust particles to become softer and tend to stick to the turbine blades. It results in modifications of the aerodynamic shape of the blades in terms of both mean shape and roughness level [20,29–32].



Figure 5. (a) Damaged compressor blades due to erosion (b) blade lost material at edges. Reprinted with permission from ref. [32]. Copyright 2015, Elsevier.

1.2.1. Erosion of Compressor Blades by Sand Particles

In order to understand the effect of sand erosion, researchers [40–43] systematically passed the sand particles into the gas turbine and measured the performance of the compressor. It was observed that the performance of the compressors decreased as the sand feed rate increased. Because of the sand erosion blade tip clearance, the surface roughness of the blade increased, the thickness of the blade trailing edge was reduced, the leading edge of the blade was blunted, and chord shortening happened [40–43]. All these resulted

in reduced efficiency of the engine and an increased consumption of fuel. Since aerospace technology needs accuracy and high safety standards, before the experiment, several authors simulated the erosion and fouling conditions and their effects on gas turbine engine performance [44–48]. Algallaf and Teixeira [44] reported degradation of the compressor performance due to eroding environments by linking the compressor aerodynamics with a thermodynamic cycle using Turbomatch, the Cranfield in-house gas turbine performance simulation software. They have uniformly increased blade surface roughness in the first two stages of a low-pressure compressor. It resulted in a significant reduction in compressor efficiency and pressure ratio. Consequently, this impacts the performance of the whole thermodynamic cycle. In another work, Brandes et al. [45] studied the degree of severity due to compressor blade erosion by taking the camber line at the blade's leading edge as the parameter. They have used flight mission simulation and a non-dimensional engine model to simulate the erosion conditions experienced by compressor blades throughout a flight mission. It was found that the high-pressure compressor experiences more erosion compared to the fan root and booster from the compressor erosion. Morini et al. [47] simulated stage-by-stage blade deterioration (i.e., fouling, mechanical damage, and erosion) in the compressor and the turbine to assess its effect on their performance. Even though the area has not decreased due to the mechanical damage, compressor efficiency and mass flow rate were reduced. Whereas, fouling and erosion caused a shift in the corrected mass flow rate and deteriorated the efficiency of the engine. Mohammadi et al. [48] used a coupled temperature-displacement finite element model to estimate the damage of a Ti6Al4V-made compressor blade exposed to alumina erodent particles. The average temperatures of the compressor's first, second, and third stages were considered 25 °C, 200 °C, and 350 °C, respectively. Following that, the simulations were conducted at all three temperatures. Their erosion results showed that the middle stage blades of the compressor at 200 $^{\circ}$ C are the most critical sites in terms of SPE damage. It was observed that there is an exponential relationship between the erosion rate and erodent velocity. The values of exponent magnitudes for the first, second, and third stages of the compressor are 2.5, 2.8571, and 2.3333, respectively. Alozie et al. [49] reported a novel "equivalent operating time (EOT)" model that considers fouling, erosion, and blade-tip wear to determine the degradation rate of the turbine engine with respect to a pre-defined reference condition. It was concluded that the model perfectly predicts the component degradation and its remaining service life. All the above simulation studies greatly helped in understanding the SPE effects on a gas turbine engine.

1.2.2. Erosion by Volcanic Ash

Airborne volcanic ash from volcanic eruptions around the world also results in damage and low safety for gas turbine-powered aircraft. In particular, Plinian eruptions can release large quantities of fine ash particles up to an altitude of 164,000 ft. Even though the large particles reach the ground rapidly due to gravity, the smaller particles stay and drift for days to weeks before reaching the ground [50]. Unlike sand, volcanic ash is usually composed of small, sharp pieces of rocks, minerals, and glass with a size in the range of 0.001 to 2 mm in diameter. The predominant component of volcanic ash is usually silica (SiO2) particles. However, it also contains other materials like Al₂O₃, TiO₂, FeO/Fe₂O₃, MgO, MnO, CaO, Na₂O, K₂O, and P₂O₅. In addition, volcanic eruptions also release sulfur dioxide (SO₂), water vapor (H₂O), carbon dioxide (CO₂), hydrofluoric acid (HF), hydrochloric acid (HCl), and ashes into the atmosphere. Acids such as HCl and HF dissolve in water and result in acid rain, whereas maximum SO_2 gradually converts into sulphuric acid (H_2SO_4) aerosols. These aerosol droplets can be absorbed by the volcanic ash particle surfaces, which are hard, extremely abrasive, and mildly corrosive [7,50-52]. When volcanic ash enters the aircraft engine, two phenomena simultaneously occur in the cold and hot sections of the engine. In a high-pressure hot turbine, particles become softer and sometimes melt due to the high gas temperatures and stick to the turbine blades. Consequently, this affects the stability of the engine and, in some cases, causes engine shutdown. It has been reported that the

volcanic ash eruption from Eyjafjallajökull, Iceland, in Europe in April 2010 resulted in estimated damage equal to USD 1.7 billion [51]. The eruption caused the interruption of air traffic for a significant period of time and damage to several aircraft components. One of the most important damages is the erosion of the compressor and turbine air foils. It results in a quick enhancement in the compressor discharge pressure and the burner pressure [52]. Similar notable incidents have also happened in the past. In 1980, a Trans-America aircraft, which was powered by an Allison T-56 turboprop, encountered an ash cloud from the eruption of Mt. St. Helens. After spending 3 to 4 min in the ash cloud, the aircraft lost two of its four engines [52]. Similarly, on 24 June 1982, British Airways flight BA009, a Boeing 747 aircraft powered by Rolls-Royce RB 211-524 engines, faced an ash cloud when flying over Indonesia from the erupting Mt. Galgunggung volcano. The aircraft travelled through the ash cloud for approximately 7 mins. As a result, all four engines of the aircraft were lost in rapid succession. However, the aircraft landed safely in Jakarta [53]. Similar incidents were faced by a Boeing 747 aircraft on July 13, 1982, and a Boeing 747 aircraft with General Electric CF6 engines in 1989 [54]. Even though no crashes were attributed to the fly ash, an average of three aircraft per year encountered volcanic ash between 1973 and 2003 [50]. Giehl et al. [55] simulated the effect of fly ash on the turbine blades and found that the chemical composition and crystal/glass ratio hold strong control over the formation of deposits on turbine blades. Vogel et al. [46] reported fluid dynamics simulations, including a Lagrangian particle-tracking algorithm, to study the volcanic ash particles and fan interactions using a high-bypass gas turbine engine design. It was concluded that the interactions between the particles and the fan redirected the particles from the engine core section into the bypass stream tube due to centrifugal effects by the fan. The interaction between fan and particles increased with increasing fan speed and volcanic ash particle size.

1.3. Overview of Solid Particle Erosion Studies and Possible Solutions

It is clear from the above discussion that solving the SPE is necessary to achieve efficient gas turbine engines with low emissions and minimum maintenance costs. Some of the initial experimental studies around the year 1960 were focused on understanding the erosion mechanism in brittle and ductile materials [3,4,39]. In 1960, Finnie [3] reported the mechanism of erosion for ductile and brittle materials with respect to the angle and speed of the eroded particles. Since then, after thoroughly analyzing the mechanisms of erosion, the solutions to overcome this problem have gradually evolved with time. Utilizing inlet particle separators and erosion-resistant coatings are the two major solutions developed over the last few decades. The inlet particle separators use different mechanisms, such as inertial impaction, diffusion, interception, and sieving, to remove the erodent particles from the air [56]. The inertial impaction removes the particles larger than 1 μ m in size, and the diffusion is effective in removing very small particles, typically less than 0.5 μ m in diameter. Whereas, through interception, medium-sized particles that are not large enough to get separated from the flow path due to inertia or not small enough to get diffused are removed. Sieving is a simple technique where the filter fibers are smaller than the particles. In addition to these techniques, electrostatic charge is also used to trap dust particles ranging from sizes of 0.01 to 1 μ m. Different filters will be selected based on the environment and the contaminants present in the ambient air, as the temperature and cleanliness of the inlet air are key to maintaining a higher efficiency of the gas turbine [56,57]. However, filtration techniques need design modifications, which result in cost increases and a time-consuming process [58]. Therefore, a lot of time and energy has gone into the development of erosionresistant coatings. Various kinds of erosion-resistant materials and coating architectures have been developed, studied, and used since the 1960s [58,59]. The protective coatings developed in the last four decades can be classified into three generations: (1) singlephase nitride and carbide coatings (1st generation); (2) metal/ceramic thick multi-layered coatings (2nd generation); and (3) nanocomposite and nanolayered multi-layered coatings (3rd generation). A detailed discussion on these coatings will be given in later sections. However, thermal barrier coatings (TBC), which were developed for high temperature (≥ 1000 °C) applications to protect the Ni-based superalloys in turbines from the high temperature airflow damage, will not be covered in this review. Briefly, the requirements for TBC coatings are a high melting point, a large thermal expansion coefficient, very low thermal conductivity, superb damage tolerance, and moderate mechanical properties [2,27]. The typical coating materials used in TBC are complex dielectric oxides with extremely low thermal conductivity [27], such as ytrria partially stabilized zirconia, i.e., ~8 wt% Y₂O₃ (8YSZ) [2,27]. Based on the compressor stage, the temperature inside the compressor varies from 100 to 600 °C. Similarly, the temperature inside the turbine can be ≥ 1000 °C. So, it is important to have a good understanding of the high-temperature erosion process. Even though there are a large number of studies on erosion, the majority of them are conducted at room temperature (RT). Following that, except for a few, most of the previous review articles discussed erosion studies conducted at RT only.

In 2006, Manish Roy [30] reviewed the SPE of metals and alloys at elevated temperatures. The review mainly focused on the oxidation of metals and their effect on erosion at elevated temperatures. In addition, the author also reviewed high-temperature erosion testing methods and concluded that the testing conditions will have a great influence on the results. However, in later studies, it was also revealed that the bonding nature, mechanical properties, and structure of the substrate materials play a great role in high temperature erosion [60]. In addition, the review did not discuss high-temperature erosion studies conducted on ceramic materials and coatings. Afterwards, in 2019, Suman et al. [61] reviewed fouling in gas turbines due to airborne contaminants. The authors discussed several methods to model particle sticking. However, the authors opined that due to a lack of precise characterization of the dust particles, the models could not predict the growth of solid deposits in gas turbines accurately. In 2014, Bousser et al. [20] reviewed mechanical characteristics such as hardness, elastic modulus, and fracture toughness of the materials in view of the SPE. The review article discussed SPE mechanisms for ductile and brittle materials, including the transition between ductile and brittle mechanisms. The authors also discussed the effect of erodent particle properties such as size, shape, and speed on SPE behavior. In the same article, nitride and carbide-based protective coatings against SPE were also reviewed. Here, the high-temperature SPE studies were not discussed. In another interesting article, Alqallaf et al. [34] published a review on SPE mechanisms (ductile, brittle, and ductile to brittle transition mechanisms) and the effect of erodent particle properties and protective coatings on erosion. However, there is no detailed discussion on high-temperature erosion studies. So, a detailed review of high-temperature erosion studies on aerospace materials will help the scientific community understand the performance of gas turbine engines.

In the present review, SPE studies conducted at high temperatures on ductile and brittle materials will be reviewed. In addition, all three-generation erosion-resistant coatings will be discussed. However, the scope of the review will be primarily focused on aerospace materials and protective coatings for compressor blades only. Models and mechanisms published in high-temperature erosion studies will be discussed. Further, parameters that influence high-temperature erosion, such as oxidation, mechanical properties, and molecular bonding, will be discussed in detail. After this, state-of-the-art coatings prepared by physical vapor deposition (PVD) techniques for SPE will be presented.

2. Temperature-Dependent Solid Particle Erosion of Metals and Ceramics

Even though the number of SPE studies conducted at high temperatures is relatively low, there are a considerable number of studies available to understand the SPE phenomena with increasing temperature. Metal alloys, such as Ti6Al4V, Inconel, aluminum alloys, and stainless steel (SS), were well studied for their high-temperature SPE properties as these materials are extensively used in the aerospace sector. Apart from these metals, cermet and other ceramic materials were also studied for their high-temperature SPE properties because of their usage in harsh environments.

2.1. Ductile or Metallic Materials and High-Temperature SPE Models

During SPE, solid particles of 10 to 300 μ m in size with sufficient kinetic energy collide with the substrate and transfer their energy fully or partially to the substrate. The transferred energy then breaks the bonds in the substrate, creating cracks in the case of brittle substrates and plastic cutting in the case of metallic substrates. Overall, depending on the substrate material, the erosion process involves different material removal procedures, such as cutting, fatigue, brittle fracture, and melting, as mentioned in the previous section (Figure 4). In addition, as the temperature increases (e.g., from RT to 800 °C), the properties of the metals get affected. For example, with increasing temperature, metals' ductile nature increases, electrical resistance increases, thermal expansion takes place, in some cases oxidation occurs on the surface, etc. [62–65]. So, these continuously varying properties with increasing temperature play a significant role in the high-temperature SPE process.

Some of the initial studies on the effect of temperature on SPE were reported by Gat and Tabakoff in their 1978 publication [66]. They have studied the SPE of alloys (Ti6Al4V, aluminum 2024, SS410) and pure metals (W, Ta, Pb) in the temperature range of RT to 200 °C. It was observed that at an impinging angle of 20°, SPE decreased with increasing temperature for all three alloys. In the case of a 60° impinging angle, erosion decreased for the SS and Ti6Al4V alloys, while the aluminum alloy showed an increase in erosion with increasing temperature. In the case of normal impingement, Ti6Al4V showed a decrease in erosion, while the other two alloys showed an increase in erosion with increasing temperature. So, Ti6Al4V alloy showed decreased erosion rate with increasing temperature for all three impinging angles (Figure 6) [66]. Similarly, pure metal W showed decreased erosion with increasing temperature at 60° and 90° impinging angles. In contrast, the erosion of Ta showed temperature independence, while the erosion of Pb exhibited increased with increasing temperature [66]. While explaining these results, the authors pointed out that the data was not available to fully understand the observations. However, they suggested that all the factors influential on the temperature-dependent SPE can be classified into two groups. The first group consists of the factors that aid the erodent particles in removing the material with increasing temperature (viz., mechanical strength, Young's modulus, and work hardening). The other group consists of the factors that resist erosion (viz., ductility and rate of dynamic recovery). So, the probable explanation given by the authors was that, with increasing temperature, these two groups play a crucial role in erosion. If the first group of factors dominates, the erosion increases with increasing temperature. While the erosion decreases if the second group of factors dominates. However, it did not explain the angle-dependent SPE nature with increasing temperature observed by Gat and Tabakoff [66]. In 1984, Sundararajan [67] took high-temperature erosion results from previous publications [66,68–70] and classified them into three groups. They are: (1) a decrease in SPE with increasing temperature [e.g., SS310 (normal impact), Al 1100 (oblique impact), SS304 (oblique impact), and Ti6Al4V (normal impact)]; (2) temperature-independent SPE [e.g., Al 1100 (normal impact), SS310 (oblique impact), and Pb (oblique impact)]; (3) increase in SPE with increasing temperature [e.g., Pb (normal impact), Al2024 (normal impact), SS304 stainless steel (normal impact), and Ti6Al4V (oblique impact)]. The author suggests that the localization model can predict the temperature dependence of the erosion in a qualitative fashion. The model suggests a ratio of the dimensionless erosion rates (E) of material at two temperatures (namely, T and $T_o (T > T_o)$) can be related as follows:

$$\frac{E(T)}{E(T_o)} = \frac{e^{\{(\frac{0.2S}{T_m})(T-T_o)\}}}{\left\{\frac{n(T)}{n(T_o)}\right\}^{1.05}}$$
(1)

where T_m is the melting point of the eroding material, n is the strain hardening exponent, and S is the flow stress. Similarly, Tabakoff [71] in his 1990 paper, suggested an equation for erosion at high temperatures based on the dependency of material yield strength on temperature.



Figure 6. Erosion rate of Ti6Al4V with increasing temperature for impinging angle: (**a**) 20° (**b**) 60° (**c**) 90° . Reprinted with permission from ref. [66]. Copyright 1978, Elsevier.

That is,

$$\operatorname{Erosion} = \left[\left(\frac{Y_{SRT}}{Y_S} \right)^a + \left(\frac{Y_S}{Y_{SRT}} \right) - 1 \right] \times \left[\left(\frac{V_1}{100} \right)^n \cos^n \beta_1 \left(1 - e_T^2 \right) + B \left(\frac{V_1}{100} \right)^m \sin^m \beta_1 \left(1 - e_N^2 \right) \right]$$
(2)

where, *A*, *a*, *n*, *B*, and m are empirical constants, Y_s = material yield strength at the operating temperature, Y_{SRT} = material yield strength at the ambient temperature, V_1 = particle impingement velocity, β_1 = particle impingement angle, e_N = normal restitution ratio, e_T = tangential restitution ratio.

Though the above equations provide some insight, erosion at high temperatures is a complex phenomenon. These models were given based on the experimental data available at the time. For example, Sundararajan [67] considered the erosion rate trend of SS, Ti6Al4V, Al, etc. with increasing temperature to formulate his model. As discussed above, the trend is not the same for all the impinging angles. These results were published before 1984. However, in a recent publication [60], the erosion rates of SS, Ti6Al4V, and Al decreased with increasing temperature for all the impinging angles. So, the proposed equations had limitations in explaining the observed erosion phenomena with increasing temperature.

2.1.1. Erosion Rate Trend with Increasing Temperature

Table 1 depicts the SPE rate trend of different alloys and pure metals with increasing temperature along with the SPE experimental conditions. Ti6Al4V is one of the most widely used alloys in the aerospace sector. The solid particle erosion properties of this alloy with increasing temperature were studied by different authors [60,66,72,73]. Except in one reference [72], in all other reports, the erosion rate decreased with increasing temperature for all the impinging angles (Table 1). Similarly, the erosion rate of Inconel, another aerospace alloy, was tested with increasing temperature [74,75]. The erosion rate decreased with increasing temperature at a 90° impinging angle (Table 1). A widely used SS304 alloy was also tested with increasing temperature for SPE studies [60,76]. The erosion rate of SS304 decreased with increasing temperature for 45, 60, and 90° impinging angles [60,76]. In contrast to the above reports, Tabakoff in 1989 reported [72] an increasing erosion rate for Ti6Al4V and Inconel alloys with increasing temperature (Table 1). This is because the author [72] did not conduct the erosion tests at low temperatures. After RT, the next test temperatures studied were 204 °C and 371 °C for Ti6Al4V and Inconel, respectively. This may have been the reason for not observing the downtrend in the erosion rate with increasing temperature [72]. In the case of the SS410 alloy, the erosion rate decreased for lower angle

impingements (20, 30, and 60°), and it increased for normal impingement with increasing temperature [66,72] (Table 1). Similarly, in the case of the Al2024 alloy, a few authors [66,73] observed increased erosion rates for 60 and 90° impinging angles and decreased erosion rates for a 20° impinging angle with increasing temperature (Table 1). In recent work, our group [60] studied the SPE of the Al8011 alloy and observed a decreased erosion rate with increasing temperature for 45°, 60°, and 90° impinging angles. In yet another study, pure metal W showed decreased erosion rate with increasing temperature [66,73]. On the other hand, Ta showed a temperature-independent erosion rate, and Pb showed an increasing erosion rate trend with increasing temperature is varying across the alloys and impingement angles. SS alloys of different grades showed different erosion trends with increasing temperature (Table 1). One of the reasons for this inconsistency may be due to the significant differences in the experimental conditions followed by different authors.

Singh and Sunderarajan [76] studied the SPE behavior of an annealed SS304 in the temperature range of 27 to 490 °C. The erosion rate of SS304 initially decreased until a temperature of 275 °C and then increased at high temperatures. These observations do not completely match the previously published results for SS alloy [67,71]. The authors stated that the several erosion models present at that time did not explain the observed erosion rate with increasing temperature as those models were based on RT erosion tests. This observation suggests that erosion phenomena at high temperatures need a deeper understanding. In addition, Singh and Sunderarajan [76] observed that oxidation of the eroding material did not play any crucial role even at the highest test temperature of 490 °C. They proved that the weight gain that occurred due to the oxidation is three orders of magnitude lower than the weight loss that happened due to the erosion. Even though there was no perfect matching model to explain the observed erosion phenomena, the authors used a closely matched localization model (Eqn. 1), suggesting that the material removal occurred by lip formation rather than subsurface crack-induced delamination at all the temperatures. A similar trend for Inconel was observed by Tabakoff [74], where he tested the Inconel-814 substrate at a 90° impinging angle in the temperature range of RT to 650 °C with an erodent speed of 140 m/s. The erosion rate decreased until a temperature of 260 $^{\circ}$ C and then further increased at higher temperatures. Similarly, Kim et al. [75] studied the SPE behavior of Inconel-625 material at test temperatures of 25, 250, 450, 650, and 850 °C using angular silicon carbide particles of an average size of 100 µm. The erosion rate of Inconel-625 at an impingement angle of 90° decreased until a test temperature of 450 °C and then increased at higher temperatures for 40 and 70 m/s erodent speeds. The authors attributed the decrease in erosion rate with increasing temperature to the decrease in hardness, mechanical strength, and Young's modulus of the substrate material along with an increase in ductility. However, they have suggested that a detailed analysis is needed to explain the observed erosion behavior with increasing temperature. The authors studied tensile strength and hardness with increasing temperature. Both the parameters decreased with increasing temperature. Accordingly, erosion should increase with increasing temperature, but the observed erosion behavior differs from this explanation. The authors used another correlation parameter, mechanical energy density, to understand the erosion phenomena. However, it also predicted the lowest erosion at 650 °C instead of 450 °C, differing from the real observation. Their other observation was that the erosion at RT and 850 °C occurred through a plastic cutting mechanism [75].

Substrate Material	The Initial Trend of the Erosion Rate with Increasing Temperature	Erodent Material—Size	Erodent Speed	Temperature Range	Ref.
Ti6Al4V	Decreases for 20°, 60°, and 90° impinging angles	Sand—164 µm	120 m/s	RT to 200 °C	[66]
Ti6Al4V	Decreases for 20° and 90°	Quartz—138 µm	120 m/s	RT to 400 °C	[73]
Ti6Al4V	Increases for 20° to 90°	Sand—165 µm	152 m/s	RT, 204, 371, 538 and 708 °C	[72]
Inconel-814	Decreases for 90°	-	140 m/s	RT to 650 °C	[74]
Inconel-625	Decreases for 90°	Silicon carbide—100 μm	70 m/s	RT to 850 °C	[75]
Inconel-718,	Increases for 20° to 90°	Sand—165 µm	152 m/s	RT, 350, 538 and 704 °C	[72]
Al-2024	Decreases for 20° and increasing for 60° and 90°.	Sand—164 µm	120 m/s	RT to 200 °C	[66]
Al-2024	Decreases for 20° and increasing for 90°.	Quartz—138–164 µm	120 m/s	RT to 400 °C	[73]
Ti6Al4V, Al-8011, and SS 304	Decreases for 45° , 60° , and 90° .	Alumina—50 µm	30 m/s	RT to 800 °C	[60]
SS304	Decreases for 90°	SiC	115 m/s	27 to 593 °C	[76]
SS410	Decreases for 20° , 60° and increases for 90° .	Sand—138 µm	120 m/s	RT to 200 $^{\circ}$ C	[66]
SS410,	Decreases for 30°,	SiC—250 μm	30 m/s	RT to 800 °C	[73]
SS410	Decreases for 20°, and increases for 90°.	Quartz—138 µm	120 m/s	RT to 400 $^{\circ}$ C	[73]
SS310	Decreases for 90°, no change for oblique	SiC—240 μm	30 m/s	RT to 800 °C	[68]
SS403, SS630	Increases for 30° , 90°	Alumina—1 mm	100 m/s	RT, 673 °C, and 973 °C	[77]
AISI 446SS	Increases for 30°—90°	Alumina—50 μm	40—100 m/s	350—750 °C	[78]
5Cr-1/2Mo	Decreases for 30°	SiC—250 μm	30 m/s	RT to 800 °C	[73]
W	Decreases for 60° and 90°	Sand—138 µm	134 m/s	RT to 200 °C	[66]
W	Decreasing for 90°	Quartz—138 µm	120 m/s	RT to 400 $^{\circ}$ C	[73]
Та	No change for 20°	Sand—164 µm	120 m/s	RT to 200 $^{\circ}$ C	[66]
Та	No change for 20° and 90°	Quartz—138 µm	120 m/s	RT to 400 $^\circ \text{C}$	[73]
Pb	Increases for 20° , 60° , and 90° .	Sand—164 µm	120 m/s	RT to 200 $^\circ C$	[66]
Pb	Increases for 20° and 90° .	Quartz—130 µm	120 m/s	RT to 300 °C	[73]

Table 1. Erosion rate trends for different alloys and pure metals with increasing temperature.

A few authors found an increase in erosion rate with increasing test temperatures for some metals [72,77–83]. Even though some materials showed an increased erosion rate with increasing temperatures, in selected cases, the authors did not test the materials at lower temperatures ($\leq 200 \,^{\circ}$ C) to observe the decreased erosion rate. As mentioned above, Tabakoff [72] reported an increased erosion rate for Ti6Al4V and Inconel-718 substrates by testing them at RT and 204 $^{\circ}$ C temperatures. Ham et al. [84] tested the SPE behavior of unidirectional silicon carbide continuous fiber-reinforced calcium aluminosilicate (CAS/SIC) glass-ceramic matrix using silica sand in the temperature range of 20 to 726 $^{\circ}$ C. The authors found a significant increase in wear rate from 0.16 mg/g to 0.26 mg/g with increasing temperature from 20 $^{\circ}$ C to 200 $^{\circ}$ C. However, after 200 $^{\circ}$ C, the overall wear rate remained

nearly the same, i.e., 0.26 mg/g, until 726 °C, except for 400 °C, where a maximum wear rate of 0.28 mg/g was observed. However, the report did not give a detailed reason for the observed phenomena. It was also observed that the scars became smoother and showed evidence of plasticity with increasing temperature. Shimuzu et al. [77] investigated the influence of mechanical properties on the erosion behavior of surface-treated stainless steels (SS403 and SS630) at high temperatures. These authors observed an increase in erosion rate with increasing temperature for both materials. At all the test temperatures, the materials showed maximum erosion rates at a 30° impinging angle and decreased gradually to the lowest values at 90° , suggesting the ductile nature of the metals. The erosion rate at a 90° impinging angle for both SS403 and SS630 materials is proportional to the decreasing percentage of hardness. In another paper, Shimizua et al. [79] found that the erosion rate of overlay welded material and forged material is 50% and 30% lower than the base metal of SS410 for test temperatures of 900 °C and 30° impinging angles. Here, the authors ascribed the increase in the erosion rate of the specimen to a decrease in the high-temperature hardness. At a temperature of 900 °C, the hardness of all specimens was reduced to approximately 70%. However, the erosion rate of overlay welding material was low because of its high hardness at high temperatures. Similarly, forged material restrained erosion regardless of hardness due to the refinement of the microstructure and the residual stress near the surface. Varga [80] studied the SPE behavior of four prospective metal alloys for high-temperature applications in the temperature range of 20 to 650 $^{\circ}$ C (Figure 7). In addition, the author has studied impact abrasion and high-stress abrasion. Among these, impact abrasion showed the highest contact severity per abrasive particle, followed by highstress abrasion and SPE. The wear rates increased with increasing temperature for all the alloys: FeCrC, NiCrW, CoCrW, and FeCrNbBWC (Figure 7). Kasich [81] et al. designed two different types of hard-facing alloys of FeCrC by incorporating Nb, Mo, and B to improve the performances of the materials at high temperatures. The authors explored the SPE of the materials at four different temperatures, from RT to 650 °C, at an impact velocity of 80 m/s. The erosion rate of these coatings increased with increasing test temperatures and impact angles. Granados et al. [82] studied the effect of swirling jet erosion on AISI 310SS at a temperature of 450 $^{\circ}$ C. The erosion results showed high erosive wear rates in swirling and non-swirling conditions. The erosive wear was even higher at high temperatures compared to RT testing. From the field emission scanning electron microscopy (FESEM) images, plastic deformation, ploughing, and cracking wear mechanisms were confirmed during erosion. Mishra et al. [78] reported high-temperature SPE properties of AISI 446SS material in the temperature range of 350–750 °C. The authors observed an increase in erosion rate with increasing temperature, which was attributed to a decrease in mechanical properties. They have developed a second-order polynomial model to predict the optimized parameters for the erosion rate generated by the response surface methodology. The erosion rate of AISI 446SS material was predicted using the artificial neural network technique, and the error was only 9.12% when compared to experimental values.

From the above studies, it is clear that most of the authors attributed the increased erosion rate with increasing temperature to reduced hardness, oxidation, etc. On the other hand, the authors who observed a decreased erosion rate with increasing temperature attributed the trend to an increase in the ductility, load bearing capacity, and rate of dynamic recovery. Even though the explanations for the erosion trend with increasing temperature are given suitably, most of the authors agreed that a better understanding of high-temperature erosion is required. In addition, the researchers did not follow uniform experimental conditions to compare the erosion data across the reports.



Figure 7. Wear rate of solid particle erosion testing at 80 m/s at (**a**) 30° and (**b**) 90° impact angle. Reprinted with permission from ref. [80]. Copyright 2017, Elsevier.

2.1.2. High-Temperature SPE Tests in a Corrosive Environment

Apart from the above-discussed parameters, which influence the SPE at high temperatures, SPE tests conducted in a corrosive environment along with high temperatures will result in more damage. The combination of erosion and corrosion results in accelerated erosion of the material compared to SPE alone [85–87]. Stott et al. [85] designed a novel test rig for high-temperature SPE and corrosion experiments. They have erosion-corrosion tested SS310, SS410, alloy 800HT, and Fe-2.25Cr-1Mo materials at two different temperatures, viz., 500 and 700 °C. To create a corrosive environment along with SPE, a mixture of hydrogen, hydrogen sulfide, and nitrogen was used. The environment caused sulphidation of all the materials at both temperatures. The SPE by alumina erodent particles at 30 m/s speed had only a small effect on the austenitic alloys at 500 °C when the corrosion effect was also minimal. However, at 700 °C, the damage was much more severe for SS310 than at 500 °C. The authors also observed more embedment of alumina erodent particles into the sulphide scales, particularly at 700 °C, which must be due to the increased plastic nature of the substrate. Similarly, Yu et al. [86] studied the erosion–corrosion properties of Fe₃Al, Ni, and Co-based alloys by a solid particle (SiO_2) and corrosive gas (SO_2) in the temperature range of 600 to 800 °C. They found that the erosion resistance of Fe₃Al-based alloys was superior to that of Ni and Co alloys in this temperature range. In addition, they observed that the eroded surface of the Fe_3Al -based alloy mostly showed a flaky appearance, whereas the Ni and Co alloys showed more of the ploughing effect. The authors explained the ploughing effect by considering the decrease in strength and increase in ductility of the metal materials at high temperatures. Due to this, the erodent particles penetrate the metal surface easily, thereby tearing the substrate, so that the eroded areas showed mostly the ploughing effect. Consequently, more material removal happened due to the ploughing compared to flake formation.

In some of the works, a few authors have simulated the erosion–corrosion effect at high temperatures on different materials. For example, Stack et al. [88] developed a model to create images of the surface morphologies of different materials after a high-temperature erosion–corrosion attack. The authors assumed spherical erodent particles with very high hardness and a parabolic growth rate of theoxide layer on the metal surface at high temperatures. Based on the simulation results, the erosion–corrosion process was classified into three groups. At first, the simulation results showed negligible corrosion at a temperature of 500 $^{\circ}$ C on the metal surface, suggesting the material removal process was dominated by the erosion process. Further, an increase in temperature to 700 $^{\circ}$ C showed considerable oxidation on the metal surface. However, the contact area between the erodent

particles and the surface contained oxidation and also metal surfaces, which suggested erosion–corrosion-dominated behavior. On the other hand, at a temperature of 900 °C, thick oxide layer formation occurred, and the oxide layer provided a barrier to the eroding particles, suggesting a corrosion-dominated regime. The authors also found that the increased erodent velocity increased erosion domination in the erosion–corrosion process.

2.1.3. Role of Oxidation in High-Temperature SPE

Apart from the corrosive environment, the oxidation of metals at high temperatures can play a significant role in SPE. As the gas turbines operate in an air environment at high temperatures, oxidation also plays a crucial role in the erosion process. However, at low temperatures (≤ 400 °C), the oxidation of metals does not influence the SPE [76,89,90]. Birks et al. [91] gave an illustration that shows erosion damage with increasing temperature along with oxidation of the metal surface. The instantaneous formation of the oxide layer is a result of the combined effect of erosion and oxidation (Figure 8). The thickness of the oxide layer increases with exposure time and temperature in the absence of the SPE. Whereas, during the erodent bombardment, the oxidation continued but the oxide layer thickness was reduced. It was concluded that the lower thickness values of the oxide layer do not influence the erosion process, and the erosion process continues to be the plastic cutting mechanism (Figure 8).



Figure 8. Classification of erosion–corrosion regimes in a target surface under high temperature erosion. Reprinted with permission from ref. [91]. Copyright 1994, Elsevier.

M. Roy [92] opined that the SPE of metallic materials at high temperatures depends on the interaction between erosion and oxidation. The author tried to improve the SPE resistance of Ni-based superalloys at high temperatures by tailoring strength and oxidation behavior. Ni-based superalloys are widely used in aircraft, naval, and vehicular gas turbines [89]. Several authors studied material loss by erosion and corrosion at high temperatures for metallic materials [90,93–95]. The studies indicated the presence of oxide layers along with the eroded rough surface on high-temperature erosion-tested metallic substrates [89,90]. Based on this, Roy et al. [89,90,96] classified four different interaction mechanisms between erosion and corrosion. They are: (1) metal erosion, (2) oxidation-affected erosion, (3) oxidation-controlled erosion, and (4) oxide erosion (Figure 8).

- Metal erosion: Most metal alloys do not get oxidized at low temperatures (<400 °C). In this scenario, the erosion mechanism is similar to the ambient conditions and follows a plastic cutting mechanism. Even though there is oxidation, the oxide layer should be thin, and it deforms similarly to the substrate. The erosion rate is thus higher at lower impact angles compared to normal impact. In this region, the erosion rate can be reduced for a given material if the critical strain for localization is reduced (Figure 8).
- 2. Oxidation affects erosion: When the temperature is in an intermediate range and the impact velocity and particle feed rate are moderate, an oxide layer of medium thickness is formed. During the SPE, the depth of deformation due to the impact of erodent particles goes beyond the oxide layer thickness and reaches the metallic substrate. As a result, the oxide layer is broken, and the softer metal comes out and mixes with the metal oxide flakes, forming a composite layer. Subsequent impacts thus cause the removal of composite material. Thus, this oxidation-affected erosion is converted from a ductile to a brittle erosion mechanism depending on the thickness of the oxide layer (Figure 8).
- Oxidation-controlled erosion: At higher temperatures (700 to 900 °C) with low erodent velocities and particle feed rates, the oxide layer grows to a certain critical thickness. However, the layer is brittle and not so adherent, resulting in easy spallation (Figure 8).
- 4. Oxide erosion: In the case of very high temperatures (>800 °C) during erosion testing, a thick oxide layer is formed on the metallic material's surface even at low impact velocities and particle feed rates. In this case, the deformation takes place only in the oxidized layer, and the metal underneath does not involve itself in the erosion process. As a result, material removal follows a brittle fracture mechanism. Since the adhesion of the oxide layer to the substrate is weak, the erosion rate values are usually high here.

Based on the observed experimental results for Ni-based superalloys, the erosion rate increases with increasing temperature in all the above mechanisms except for oxide erosion [95,97,98].

2.1.4. Role of Bonding Nature in High-Temperature SPE

In previous discussions, the authors used oxidation and mechanical properties to explain their erosion results with increasing temperature. Recently, our group [60] studied the temperature-dependent erosion properties of different metals (Ti6Al4V, SS 304, Al-8011, and Ti/TiN) and non-metals (Al_2O_3 , fused quartz, and Si) in the temperature range of RT to 800 °C following ASTM-G76-13 standards (erodent = angular Al_2O_3 of average size 50 μ m, erodent speed = 30 m/s and 100 m/s, angles = 45, 60, and 90°). It was observed that the erosion rates of metallic and ionic solids (viz., Ti6Al4V, SS304, Al-8011, Al₂O₃, Ti/TiN) decreased with increasing temperature and then increased at higher temperatures. In the case of Ti6Al4V and Al₂O₃, the erosion rate decreased until 600 $^{\circ}$ C, whereas for SS304, Al-8011, and Ti/TiN, the erosion rate decreased until 400 $^\circ$ C and then increased for both the set of materials at higher temperatures for a 30 m/s erodent speed. Figure 9 shows the erosion rate trend and the depth of the erosion scars for Ti6Al4V and Al_2O_3 substrates with increasing temperature [60]. It is clear that the above materials have different crystal structures, melting points, and density values [99–102], still all the above materials showed decreased erosion rates with increasing temperature. So, it infers that the above-mentioned parameters may not play a vital role in the observed erosion trend with increasing temperature. In addition, the role of oxidation was ruled out as the SS304 proved to be a good oxidation-resistant material until 870 $^\circ$ C and 925 $^\circ$ C in discontinuous and continuous services, respectively [103]. Whereas, in the case of Ti6Al4V, the oxidation starts around 500 °C and then becomes more significant only after 800 °C [104]. In the case of fused quartz and Si materials, the erosion rate increased with increasing temperature from

RT to 700 °C (Figure 10) [60]. Along with the erosion rate, scar depth and the 2-D profiles of the erosion scars for different test temperatures can be seen in Figure 10. These two materials are covalently bonded. The study concluded that, with increasing temperature, lattice vibrations and bond flexibility increase in the case of non-directional ionic or metallic bonds. As a result, the ductility and load-bearing capacity of the materials increase, and the erosion rate decreases. However, at higher temperatures, the bonds become weaker, resulting in increased erosion after a certain temperature. On the other hand, covalent bonds are strong and directional. These bonds do not become flexible with increased temperature. This is the reason why the erosion rate increases in the case of quartz and Si substrates with increasing temperatures.



Figure 9. (a) Erosion rate of Ti6Al4V with respect to temperature. (b) 2D profiles of the erosion scars of Ti6Al4V substrate at different temperatures for a 90° impinging angle. (c) Erosion rate of Al_2O_3 substrate with respect to temperature. (d) 2D profiles of the erosion scars of the Al_2O_3 substrate at different temperatures for a 90° impinging angle. Reprinted with permission from ref. [60]. Copyright 2020, AIP Publishing.

In brief, it is clear that the temperature significantly influences the erosion rate of the metals and alloys. Several authors have reported the SPE performance of different aerospace alloys and metals with increasing temperature. The majority of the authors observed an initial decrease in the erosion rate with increasing temperature for different alloys and metals (Table 1). As the temperature increases, toughness, rate of dynamic recovery, and load bearing capacity of the materials increase. This may be the reason for the initial decrease in the erosion rate. In other terms, these materials are made of ionic or metallic bonds. These bonds are flexible and non-directional. With increasing temperature, the bonds become more flexible and act like springs, which results in increased toughness and load bearing capacity of the materials. In addition to these effects, oxidation

of the metals at high temperatures also plays a role in the SPE process. However, reports suggested that until a temperature of ~ 400 °C is reached, oxidation does not significantly influence the erosion process of most of the alloys and metals. This may be the reason for observing the decreased erosion rate with increasing temperature initially. However, at higher temperatures, the surface of the metals will be significantly oxidized, changing the phase and properties of the surface. This results in oxidation-influenced erosion. In the case of the erosion process in the corrosion media, the erosion process will be accelerated due to the combined erosion–corrosion effect.



Figure 10. (a) 2D profiles of the erosion scars of fused quartz at different temperatures for 90° impinging angle. (b) Erosion rate of fused quartz substrate with respect to temperature. (c) 2D profiles of the erosion scars of Si (100) at different temperatures for 90° impinging angle. (d) Erosion rate of Si (100) substrate with respect to temperature. Reprinted with permission from ref. [60]. Copyright 2020, AIP Publishing.

2.2. Solid Particle Erosion of Brittle Materials

Several brittle materials were also tested for their high-temperature erosion studies. In order to understand the erosion process with increasing temperature high-temperatures, erosion studies on brittle materials are also discussed here. Choe et al. [105] examined the fracture toughness (R-curve) and fatigue crack growth properties of Mo-12Si-8.5B in the temperature range of RT to 1300 °C. The alloy displayed a relatively high intrinsic toughness and limited extrinsic R-curve (crack-growth) in the temperature range of 800 to 1200 °C. With increasing temperature, the toughness and resistance to fatigue-crack growth of the alloy increased. The increased toughness was ascribed to a variety of toughening mechanisms that became active at high temperatures. Importantly, crack trapping by the a-Mo phase and micro-cracking, primarily in the Mo₅SiB₂ phase, was responsible for an increase in toughness with increasing temperature. The increased toughness can result in reduced SPE. Alman

et al. [106] explored the SPE resistance properties of FeAl-WC, FeAl-TiB₂, and FeAl-TiC cermets at 25, 180, 500, and 700 °C. They ranked the materials in terms of erosion resistance performance as follows: FeAl–WC > FeAl–TiB₂ > FeAl–TiC. With increasing temperatures, the temperature erosion rate of FeAl–TiB₂ remained unchanged, whereas the same decreased for FeAl–WC and FeAl–TiC cermets. In the case of WC–Co, the erosion rate increased with increasing test temperatures. The authors ascribed the observed erosion nature with increasing temperature to the oxidation of the binder materials. Oxidation of "Co" in WC-Co caused an increase in the erosion rate with increasing temperature. In the case of FeAl-cermets, the oxidation was low at elevated temperatures; hence, the erosion rate of these cermets was constant or decreased. Though the authors used oxidation to explain the observed erosion phenomena, the reason is not sufficient to describe the observed results, such as a decrease in erosion rate with increasing temperature. In another work, Chaves et al. [107] studied the erosion-corrosion resistance of TiC-NiMo cermets at high temperatures. They opined that erosion at high temperatures in an oxidizing environment depends on the rate of oxidation, adherence, morphology, and toughness of the oxide layer formed on the surface of the material. Similarly, Hussainova et al. [108] evaluated the SPE performance of Ti(Mo)C-Ni cermets at 25, 350, and 650 °C using silica as the abrasive. They have observed a nearly constant erosion rate up to a temperature of 650 °C for the cermets with low binder content. The authors ascribed the observation to an increase in the plasticity of the cermets and the formation of the thin oxide layer with increasing temperature. In another pioneering work, Wang et al. [109] studied the SPE behavior of high-purity alumina ceramics at 800, 1100, and 1400 °C. The erosion rate remained constant until 800 °C but increased significantly after 800 °C for different impingement angles (30°, 45°, 60°, 75°, and 90°). The authors observed dominating brittle fracture mechanisms, even at elevated temperatures. The observed erosion phenomena were ascribed to a decrease in the hardness and yield strength of the alumina with increasing temperature. The drop in these parameters is more significant after 800 °C [110]. In a similar work, Li et al. [111] tested the SPE properties of SiC-Si₃N₄ composite ceramics at different temperatures from RT to 1400 °C, using sharp SiC erodent particles of diameter $325-830 \,\mu m$ (Figure 11). Initially, the erosion rate of the SiC–Si₃N₄ composite ceramic increased with increasing temperature until 800 °C and then decreased until 1400 °C for all the impinging angles (Figure 11). It is clear from the figure that the erosion rate increased with increasing impinging angles from 30° to 90° , confirming the brittle nature of the material. The authors ascribed the decrease in the erosion rate until 800 °C to a reduction in the hardness of the material matrix with increasing temperature. An increase in the erosion rate after 800 °C was attributed to oxidation of the material, which could have increased the wear resistance of the sample. In addition, at high temperatures, bonding between coarse aggregate and the matrix increases due to the expansion of the coarse aggregate, which results in improved spalling resistance and erosion resistance of materials.

In their 2015 work, Fang et al. [112] studied the SPE wear behavior of 3 mol-% yttriastabilized zirconia ceramics at different test temperatures, from RT to 1400 ° at an impact angle of 90°. Their results showed a steady increase in erosion rate with increasing temperature until 600 °C and then a rapid rise between 800 and 1200 °C. The authors attributed the increase in erosion rate to a decrease in mechanical properties, such as bending strength, fracture toughness, and elastic modulus, with increasing temperature. The erosion wear mechanism at low temperatures (below 800 °C) followed plastic deformation, and at high temperatures (>800 °C), cracking in irregular criss-cross patterns was observed.



Figure 11. Erosion rates of SiC–Si₃N₄ composite ceramic with increasing test temperature and the impingement angle. Reprinted with permission from ref. [111]. Copyright 2014, Elsevier.

3. Protective PVD Coatings for Solid Particle Erosion

3.1. First-Generation Coatings (Single-Phase Coatings)

As discussed in the previous section, alloys used in aerospace and other sectors, such as Ti6Al4V, Inconel, and SS, are susceptible to SPE [3,20,113–116]. Over the last four decades, different types of erosion-resistant coatings have been developed. Initially, singlephase nitride coatings, such as TiN, TiAlN, and CrAlTiN, were developed to protect the gas turbine compressor blades [3,24,31,117]. Jianxin et al. [118] deposited four different nitride coatings, CrN, ZrN, CrAlN, and TiAlN, on YT15 cemented carbide using the cathodic arc evaporation technique. Their erosion results showed that the coatings with Al (CrAlN and TiAlN) content performed better than SPE over CrN and TiN coatings. The authors also claimed that the higher the H^3/E^2 value of the coatings, the better the erosion resistance. In the case of TiN and CrN, material removal happened through a brittle fracture mechanism, while in the case of AlTiN and CrAIN coatings, mainly micro-cutting and cycle fatigue fracture happened. Sue and Troue [117] studied the high-temperature erosion resistance performance of TiN and ZrN coatings using 27 µm alumina particles with 120 m/s impact velocity. The authors tested the coatings at 30° and 90° impingement angles at different temperatures of 25, 316, and 538 °C. The erosion rates of both coatings increased with increasing temperature for both impinging angles. The increased erosion rate was ascribed to relaxed residual compressive stress in the coatings with increasing temperature. Tabakoff [74] tested different coatings using an in-house designed and developed high-temperature erosion test rig facility. The experimental conditions simulated the aerodynamic conditions on the blades. Along with other alloys, the author tested CVD deposited TiC, TiN, and Al_2O_3 coatings using the following test conditions: 250 μ m alumina erodent particles, temperatures from RT to 650 °C, impact velocity of 140 m/s, impinging angle of 90° (Figure 12). Based on the results, the author concluded that the temperature has a very weak effect on the erosion rate of Al_2O_3 and TiN coatings. Whereas the erosion rate of TiC initially decreased with increasing temperature until 260 °C and then increased at higher temperatures. In another work, Shanov et al. [119] deposited a diamond coating using the CVD process and studied its SPE resistance properties at different temperatures. The SPE tests were conducted at a 90° impingement angle with a

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particle velocity of 152 m/s. The erosion rate of diamond coating decreased significantly with increasing temperature, from ambient to 538 °C. The authors opined that, in order to interpret these results, additional experimental investigations such as microstructure, oxidation, and fracture mechanics were necessary. Similarly, Leithead et al. [120] tested bare 17-4PH steel V103 profile gas turbine engine axial compressor blades and those coated with either CrAlTiN or Ti_xAl_{1-x}N erosion-resistant coating. The coatings were deposited up to a thickness of 16 µm using the physical vapor deposition technique. The authors claimed that the tested experimental rig was able to create actual field conditions. Based on the results, CrAlTiN and Ti_xAl_{1-x}N-coated blades performed at least 79% and 93% better than bare blades, respectively. Our group [121] deposited an H-free single-phase diamond-like carbon (DLC) coating of thickness 6 µm on Cr/CrN-coated plasma-nitrided steel using an unfiltered cathodic vacuum arc system. Solid particle erosion tests were conducted according to ASTM-G-76-18 standards in the temperature range of 25 to 400 °C at an impinging angle of 45° . The coating showed better erosion resistance performance compared to a nitrided steel substrate at all the test temperatures. The erosion rate of the coating increased with increasing temperature. The observed erosion nature was ascribed to the covalent bonding of C atoms in the DLC sample, which are directional and strong. In addition, graphitization of the DLC sample at a test temperature of 400 °C resulted in a high erosion rate at that test temperature.



Figure 12. Variation of the coating erosion rate with the temperature at a 90° impingement angle. Reprinted with permission from ref. [74]. Copyright 1999, Elsevier.

As mentioned earlier, SPE is a complex process—different mechanical failure mechanisms are involved, and different properties of the erodent and substrate influence the erosion process. Apart from the initial and final conditions, experiments cannot give more details on the erosion process. In this scenario, simulation studies can shed some light on what happens during the erosion process. Few authors have simulated the erosion process in brittle and ductile materials. For instance, Griffin et al. [122] used ABAQUS/Explicit analysis and simulated multiple impacts by silica particles at impinging angles varying from 20° to 90°. Two failure models (viz., a tensile failure model for the brittle material and a shear failure model for the ductile material) were considered. In the case of a 20° impinging angle, signs of crack formation were observed after five impacts, but no material loss was seen. For Al_2O_3 , the rate of erosion increases with increasing impinging angles, suggesting its brittle nature. In another work, Zheng et al. [123] simulated erosion caused by continuous particle impingement by considering material damping and the elastic-plastic behavior of the target material. It was observed that the first collision caused the maximum plastic deformation, which decreased with an increase in collision numbers. Similarly, erosion scar width and depth initially increased sharply and then gradually increased with increasing collision numbers.

To summarize, the hard single-phase first-generation coatings show excellent SPE resistance for glancing impinging angles but exhibit poor SPE resistance for normal impinging angles due to their low toughness [20,124]. In addition, the deposition of single-phase nitride coatings of thickness above 5 μ m is difficult due to the high internal stress. High internal stress leads to low adhesion and poor erosion resistance performance. These drawbacks led to the development of second-generation metal/ceramic multilayered coatings.

3.2. Second-Generation Coatings (Metal/Ceramic Multilayer Coatings)

In the metal/ceramic multilayer design, the soft metal layer reduces the internal stress, allowing the growth of thicker and more adhesive coatings with high toughness compared to single-phase nitride coatings [24,125]. In addition, during SPE, cracks generated in the hard layer can be stopped by the immediate next soft metal layer. Figure 13 shows crack deflection at the interface between layers, ductile interlayer ligament bridging and crack tip blunting due to soft metal nano-sized layers [125]. Kolkman [126] reported the mean coating stress of a $1/1 \,\mu m$ (Ti/TiN) multilayer coating was 670 MPa, whereas the same for thick single-phase TiN was found to be 2570 MPa. Over the years, researchers have investigated different materials like Ti/TiN [24,126–129], Cr/CrN [130], W/WN [131], and Ti/TiSiCN [132] to develop soft/hard multilayer coatings for SPE resistance. The authors varied the bilayer thickness from around 20 nm to 3 μ m. Based on the above studies, it was well established that the multilayer design is superior to the single-phase nitride coatings in resisting the SPE. In addition to the multilayer design, optimized thickness values for hard and soft layers provide superior erosion resistance performance. For example, the erosion rate of Cr/CrN multilayers varied significantly for different thickness values of Cr and CrN. The erosion rate of Cr/CrN film was 150 times lower when the ratio between the thickness values of Cr and CrN was 0.8, compared to the Cr/CrN coating with the same ratio of 0.15 [130]. Even though different materials were studied, the majority of the recent papers reported Ti/TiN multilayered coatings for SPE resistance [24,126–129]. For instance, Borawski et al. [133] deposited erosion-resistant Ti/TiN multilayers of thickness $25 \ \mu m$ for the protection of AM355 steel turbine components. The authors varied the number of layers as 2, 8, and 32 with three different thickness ratios: 79:1, 19:1, and 3:1. The SPE tests were conducted using glass beads, quartz, and alumina media with erodent velocities ranging from 75 to 180 m/s. The authors found that the erosion performance was highly dependent upon the TiN/Ti coating design architecture and the erosion conditions. Coatings with two layers showed good erosion performance against the alumina erodent, whereas coatings with 32 layers performed better against the glass beads. However, very few high-temperature SPE studies are available in the literature on multilayer PVD coatings. In their work, Wang et al. [134] studied the high-temperature SPE properties of arc sprayed FeTi/CrB and SH-SAM (stainless amorphous metal) coatings at an impact angle of 30° and 90° in the temperature range of RT to 650 °C. The composite coating FeTi/CrB performed better than the SH-SAM coating at all test temperatures for both impinging angles. The erosion rates of both coatings increased with increasing temperature and then decreased after around 400 °C. The better performance of the FeTi/CrB composite coating was ascribed to its good toughness and high spalling resistance because of the hard phase at elevated temperatures.



Figure 13. Schematic representation of toughening mechanisms in multilayer films: (1) crack deflection, (2) ductile interlayer ligament bridging, and (3) crack tip blunting due to nanoplasticity at the interlayer. Reprinted with permission from ref. [125]. Copyright 2005, Elsevier.

In addition to the experimental studies, suitable erosion-resistant coatings were predicted using simulation studies. Bielawski and Beres [135] used a finite element methodology and simulated single particle impact on multilayer coatings to probe the tensile stresses on the surfaces of multilayered coatings. It was observed that the Ti interlayer of thickness 0.25 and 0.50 μ m lowers the tensile peak stress in the coating surface by 1.8% and 3.3%, respectively. The authors concluded that the peak stress on the surface of a bilayer coating system can be considerably lowered by using high-modulus bottom layers and low-modulus top layers. In another important work, Hassani et al. [136] used advanced finite element methods to predict the best erosion-resistant coating by studying the erosion mechanisms and by taking surface stress reduction as an erosion performance criterion. The authors studied single and multilayered TiN and nanocomposite nc-TiN/a-SiN₁₃ and nc-TiCN/a-SiCN coatings on Ti6Al4V and SS substrates. In the case of single-layer coating, their calculations suggested a combination of low modulus and relatively high thickness coatings can reduce the local stress and, hence, improve erosion resistance. The model also suggested that multilayer and superhard nanocomposite coatings perform better than single-layer coatings against SPE. The calculations predicted two multilayer configurations: (4 μ m TiN/4 μ m nc-TiN/a-SiN_{1.3}) and (2 μ m nc-TiN/a-SiN_{1.3}/2 μ m TiN/4 μ m nc-TiN/a-SiN_{1.3}) as the best coatings to become high-performing erosion-resistant coatings. In another publication, Hassani et al. [137] used finite element methodology to predict superior erosion-resistant coatings with appropriate stress management. The calculations suggested that the elastic modulus should increase across the coating thickness with a maximum value near the coating/substrate interface for a better erosion-resistant coating. The authors also suggested an optimal coating configuration, which consists of four layers with a specific Young's modulus (100 to 600 GPa) and specific residual stresses (-1 to -6 GPa), along with the coating depth (Figure 14). Even though the multi-layered coatings with micrometer-sized bi-layers performed better for all the impinging angles, the efficiency of such coatings was not satisfactory. In this scenario, due to rapid advances in nanotechnology, third-generation nanocoatings have emerged in recent years.

3.3. Third-Generation Coatings (Metal-Ceramic Nanocomposite Coatings and Nanolayered-Multilayer Coatings)

Over the last two decades, nanotechnology has revolutionized several sectors such as medicine, electronics, optics, data storage, energy, security, manufacturing, etc. By reducing the grain size to less than 100 nm, the material properties at the nanoscale (1–100 nm) differ significantly from their bulk counterparts. By utilizing this fact, the properties of

the materials can be manipulated according to the requirements. Following this concept, researchers have developed a large number of erosion-resistant nanocoatings, which were found to be superior to the previously developed single-phase and thick multi-layer coatings [60,113,134]. The mechanical properties of the films, such as hardness, increase when the grain size is reduced to below 10 nm. Mechanisms, namely, coherency strain, Hall–Petch strengthening, etc., have also been used to explain the hardness enhancement in the nanolayered multilayer coatings [138,139]. In his pioneering work in 1951, Hall postulated that as the grain size reduces, the hardness of the material increases significantly [139]. The performance of metal-ceramic nanocomposite coatings and nanolayered multilayered coatings with grain sizes less than 10 nm was found to be better than that of thick multi-layered coatings against SPE [113,134,140]. This improved SPE performance on nanolayered multilayer coatings can be attributed to different mechanisms. Firstly, a high number of interfaces can be created between different shear moduli materials (e.g., Ti: 45 GPa; TiN: 250 GPa), and during erosion, these interfaces restrict crack propagation effectively. Secondly, during impact, nano-sized grains can adjust among themselves to reduce the stress generated due to collision [113,141]. Thirdly, adjacent grains differ in orientation, thus, dislocations need very high energy to change their direction and move into the adjacent grain.



Figure 14. Optimal distribution of coating properties for enhanced erosion resistance for a four-layer system. Reprinted with permission from ref. [137]. Copyright 2010, Elsevier.

Swaminathan et al. [113] in Southwest Research Institute reported the development of hard erosion-resistant nanostructured titanium silicon carbonitride (TiSiCN), titanium nitride (TiN), and multilayered nanocoatings for steam and gas turbines using the plasmaenhanced magnetron sputtering process (PEMS). The optimized coatings performed 10 to 100 times better than the bare Ti6Al4V substrates (Figure 15). The authors prepared 18 samples and optimized three coatings containing grain sizes of 5.6 nm, 6.9 nm, and 7.2 nm; these grain sizes were lower compared to those of the remaining 15 samples. The authors attributed the better erosion resistance of the coatings to the extremely fine grain size. The performance of these coatings was an order of magnitude higher than monolithic TiN coatings. The same group [140,142,143] reported deposition of TiSiCN, TiAlVSiCN, and ZrSiCN nanocomposite coatings with grain sizes of ~7 nm using the PEMS technique. The nanocomposite coating showed 100 times higher erosion resistance compared to the uncoated material. They tested their nanocomposite coatings in another independent laboratory at temperatures up to 426 °C to match the gas turbine compressor environments and found that the erosion was reduced by 50–90%. Figure 16 shows transmission electron microscopy (TEM) images (Selected Area Electron Diffraction (SAED) and HRTEM) of the TiSiCN and TiAlVSiCN nanocomposite coatings with Si contents of 2.6 at.% and 1.7 at.%, respectively [140]. A dense and fine polycrystalline grain structure can be seen in the TiSiCN coating (Figure 16a). Whereas the TiAlVSiCN coating contains a relatively bigger columnar grain structure with column widths between 20–30 nm (Figure 16b). In Figure 16c,d, clear lattice fringe patterns of TiSiCN and TiAlVSiCN coatings can be observed. In Figure 16c, fine nanocrystallites of sizes 5–10 nm surrounded by thin amorphous phases can be clearly observed for TiSiCN coating [140]. In another work, Presby [144] developed melt-infiltrated (MI) silicon carbide fiber-reinforced silicon carbide (SiC/SiC) ceramic matrix composites and an a-SiC monolithic coating. Erosion tests were conducted using 150 μm Al₂O₃ erodent particles at 1200 °C with impact velocities from 100 to 200 m/s and the angle of impingement from 30 to 90° . It was found that the performance of the a-SiC coating against SPE is two orders of magnitude better compared to the MI SiC/SiC coating. The poor performance of MI SiC/SiC coatings was ascribed to a weak fiber-matrix interface in the MI SiC/SiC coating. In a very recent study, Maity et al. [145] developed erosion-resistant spark plasma sintered NbB2-ZrO2 composite and monolithic NbB2. The SPE tests were conducted at temperatures of 25, 400, and 800 $^{\circ}$ C using 50 μ m Al₂O₃ erodent particles impinged at a 90° angle with a velocity of 50 m/s. For both materials, the erosion rate decreased with increasing temperature. The authors ascribed this phenomenon to the relaxation of residual stress at high temperatures, the accumulation of dislocations near the crater region, and the shot peening phenomena. The performance of the composite material against erosion was two orders of magnitude higher compared to monolithic NbB₂. It has been reported that at RT, the material removal process is due to the brittle fracture mechanism, whereas at higher temperatures, intergranular cracking, rolling/sliding wear, and dislocation plasticity were observed.



Figure 15. Comparison of erosion rates of commercial TiN coating by CAPVD and the advanced nano-composite coatings by PEMS. Reprinted with permission from ref. [113]. Copyright 2010, ASME.



Figure 16. Cross-sectional TEM images, SAED patterns, and HRTEM images of the TiSiCN (2.6 at.% Si) (**a**,**c**); TiAlVSiCN (1.7 at.% Si) (**b**,**d**). Reprinted with permission from ref. [140]. Copyright 2018, Elsevier.

Apart from SPE, corrosion caused by neutral sea salt and humid air also plays a crucial role in damaging compressor blades. In comparison with nanocomposite coatings, nanolayered multilayer coatings can prevent corrosion effectively because the corrosion medium can penetrate through the grain boundaries in the case of nanocomposite coatings. Whereas parallel interfaces in multilayers restrict the corrosion media from passing through the coating [146,147]. Nitride/nitride nanolayered multi-layered coatings were developed for erosion–corrosion resistance [148,149]. These coatings show good corrosion resistance but failed to protect the substrate against SPE due to the high internal stress and poor adhesion. Wolfe et al. [148] deposited erosion-resistant nanolayered structures consisting of TiN-rich and CrN-rich layers with periodicity, ranging from 10.5 to 21.0 nm onto AM355 stainless steel using a multi-source cathodic arc system. The micro-hardness of deposited coatings ranged from 1700–2800 kg/mm². Erosion tests were conducted using glass beads and alumina particles at an erodent velocity of 150 m/s. Films with a high number of nanolayer interfaces showed poor erosion resistance in the glass bead tests. This may be because both the layers (TiN and CrN) are hard nitride layers, which will result in high interfacial stress and low toughness. Supporting this, coatings with a high number of interfaces showed poor adhesion. Among all the coatings deposited, coatings with a higher CrN content and the lowest number of interfacial sizes showed the best erosion resistance performance. From this study, it is clear that a high number of interfaces between two hard nitride layers do not work well against SPE.

Unlike the above studies, some authors used metal and ceramic materials to develop nanolayered, multilayered coatings. Feuerstein and Kleyman [24] developed an erosionresistant TiN/TiN_{1-x} nanomultilayer coating of thicknesses between 25 and 50 μ m using the cathodic arc PVD technique for the protection of helicopter compressor blades and vanes (Figure 17). The samples were tested according to ASTM standard G-76 using 50 μ m alumina erodent particles. The erosion resistance performance of the nanomultilayer coating was much better than monolithic TiN. For a 90° impinging angle, the optimized coating showed an erosion rate of $\sim 2 \times 10^{-3}$ mm³/g. The authors ascribed the better erosion resistance of the coating to significantly smaller crystallites, which resulted in good hardness and toughness. Our group [150] sputter-deposited Ti/TiN nanomultilayer coatings on a Ti6Al4V substrate with a bilayer thickness of 7.5–115 nm and a total thickness of 7–10 μ m. The coatings were erosion tested according to ASTM G76-13 standards at 400 $^{\circ}$ C temperature for 30, 45, 60, and 90° impinging angles. Among all the samples studied, the coating with a bilayer thickness of 7.5 nm showed very good SPE resistance, and its average performance was 15 times better than the Ti6Al4V substrate. The better erosion resistance performance of the film has been attributed to its ultra-small grain size, a large number of interfaces, and the reduced stress in the film caused by annealing. However, N diffusion into ultrathin Ti (~3.5 nm) layers resulted in high stress for the coating [150]. In order to reduce the stress and improve the erosion-resistant performance of the film, thick metallic stress absorbing (SA) layers were introduced into the film (Figure 18) [151]. It has been demonstrated that the SA layer controlled the internal stress and also balanced the brittle and ductile nature of the coating. The average erosion rate at 400 ⁰C of Ti6Al4V substrate was 74, 13, and 12 times higher than the optimized coating for the erodent speeds of 30, 60, and 100 m/s, respectively. The improvement in erosion resistance was ascribed to SA layers and ultra-thin bilayers [151]. It has been widely reported that at higher temperatures $(\geq 450 \text{ °C})$, TiAlN is chemically more stable than TiN [20]. To take advantage of TiAlN over TiN, our group developed a TiAl/TiAlN nanolayered multilayered coating and tested it according to the ASTM-G76-13 standards at impinging angles: 90°, 60°, 45°, and 30° at a test temperature of 400 °C with increasing erodent velocity from 30 m/s to 90 m/s using 50 μ m alumina erodent particles [152]. The erosion rate of TiAl/TiAlN coating at 30 m/s was 0.025×10^{-3} , 0.12×10^{-3} , 0.035×10^{-3} , and 0.03×10^{-3} mm³/g for the angles 30°, 45°, 60°, and 90°, respectively. Apart from SPE resistance, the coating showed very good corrosion resistance. Until an erodent speed of 50 m/s, TiAl/TiAlN coating erosion resistance performance was better than that of Ti/TiN coating. However, at higher erodent speeds of 60, 70, 80, and 90 m/s, the Ti/TiN coating performed better than that of TiAl/TiAlN. The erosion rate increased almost linearly with increasing erodent speed. Figure 19 shows the temperature-dependent erosion rate of Ti/TiN for a 90° impinging angle with a 30 m/s erodent speed [60]. The erosion rate decreased with increasing temperature until 400 °C and then increased at temperatures higher than 400 °C. An initial decrease in the erosion rate can be ascribed to increasing toughness with increasing temperature. However, at high temperatures, bond strength decreases, resulting in higher erosion rates.

3.4. SPE Resistant Coatings with Energy-Absorbing Nano-Porous Metal Layers

As mentioned in the previous section, our group has developed an erosion-resistant Ti/TiN nanolayered multi-layered coating with stress-absorbing layers (Figure 18). The erosion performance of this coating was further improved by introducing porous metal layers [153]. Metallic foams such as Ti, Al, and Mg do have a higher energy absorption capacity compared to their dense counterparts [154–160]. So, these metallic foam materials are used in car crash protection and armor. Inspired by this, in the above-mentioned coating, Ti SA layers have been replaced with nanoporous Ti layers (Figure 20) [153]. Deposition of the porous Ti metal layer was realized by following Thornton's grain microstructure map for PVD coatings [161,162]. Two different ultra-thin Ti/TiN multilayered coatings with porous (Ti/TIN-P) and dense (Ti/TIN-D) Ti metal layers (~320 nm) were sputter deposited on Ti6Al4V substrates. The SPE tests were conducted according to ASTM-G76-13

at impinging angles of 30 to 90°, erodent speeds of 30 to 100 m/s, and temperatures of 25 to 700 °C. The average erosion resistance performance of Ti/TiN-P coating with porous Ti layers was ~44 times better than Ti6Al4V substrate and ~3.3 times better than Ti/TIN-D coating with dense Ti layers for an erodent speed of 100 m/s (Figure 21a). The better performance of the Ti/TiN coating with porous layers was ascribed to the higher energy absorption capacity of porous Ti layers. In order to validate the experimental results, finite element simulations were made to simulate the experimental conditions. The finite element simulations showed that the energy absorption rate of porous Ti metal layers is higher compared to dense Ti layers at erodent speeds of 30 to 100 m/s. The energy absorption capacity of porous Ti layers improved with increasing erodent speed [153]. Similarly, the erosion-resistance performance of Ti/TiN-P coating is much better at higher erodent speeds (≥50 m/s) in comparison with Ti/TiN-D coating (Figure 21b). Temperature-dependent SPE studies were also conducted on the Ti/TiN-P coating, and the erosion rate of it decreased with increasing temperature and then increased at higher temperatures due to its bonding nature [153]. In this work, the combination of nano sized grain size and porous layers resulted in improved erosion resistance of the coating. This is the first study in which porous metal layers are used in the erosion-resistant coating. Further studies in this direction may lead to the development of efficient erosion-resistant coatings.



Figure 17. Coating architecture of a $\text{TiN}/\text{TiN}_{1-2}$ 4k Type IITM multilayer coating system for small and medium particle erosion protection. Reprinted with permission from ref. [24]. Copyright 2009, Elsevier.

3.5. Notable New Generation Erosion Resistant Coatings

As discussed in the previous sections, several different types of coatings have been developed for SPE resistance. With time, the erosion-resistant coatings evolved toward better efficiency. In this section, we would like to mention state-of-the-art PVD coatings for compressor blades to the best of our knowledge. As the erosion-resistant coating needs higher thicknesses along with good mechanical properties, several organizations/companies used PVD deposition techniques that give higher growth rates, such as plasma enhanced magnetron sputtering and cathodic arc, to deposit these coatings [24,113,142,163]. Higher growth rate PVD techniques along with low-cost precursor materials can save time and production costs. Some of the thick notable/commercial erosion-resistant coatings are tabulated in Table 2. A detailed discussion on some of these coatings was given in the previous section. The common features of these coatings are their high thickness ($\geq 9 \mu m$), high number of interfaces, lower grain size, metal/ceramic multilayer, or composite design, balanced ductile and brittle nature, lower internal stress, and good adhesion.



Figure 18. Cross-sectional FESEM image of Ti/TiN. The high resolution image in the inset shows the Ti/TiN multilayers. Reprinted with permission from ref. [115]. Copyright 2019, Elsevier.



Figure 19. Erosion rate of a Ti/TiN film with respect to temperature for a 90° impinging angle. Reprinted with permission from ref. [60]. Copyright 2020, AIP Publishing.



Figure 20. (a) Cross-sectional FESEM image of Ti/TiN-P coating, (b) porous Ti layer, and (c) schematic of Ti/TiN ultra-thin multilayers. Reprinted with permission from ref. [153]. Copyright 2021, Elsevier.



Figure 21. (a) Angle-dependent erosion rates of Ti/TiN-P and Ti/TiN-D coatings at conditions: temperature 400 °C and erodent speed 100 m/s. (b) Erodent speed dependent erosion rates of Ti/TiN-P and Ti/TiN-D coatings at conditions: impinging angle 90° and temperature 400 °C. Reprinted with permission from ref. [153]. Copyright 2021, Elsevier.

Table 2. Notable/commercial state-of-the-art solid particle erosion re	esistant coatings.
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S. No.	Coating	Total Thickness	Deposition Method	Hardness	Erosion Rate	Ref.
1	Metal (Ta, Nb, Ti, etc.) /Ceramic (TiAlCrC, TiAlN, B ₄ C, etc.) multilayers. Bi-layer thickness is in few microns	15 to 25 μm	Magnetron sputtering, Cathodic arc	≥19 GPa	15–20 μm thickness loss for 200 gr of erodent (50 μm alumina).	[164]
2	TiN/TiN _{1-x} (0.2/1 μm) multilayer coating	15 to 25 μm	Cathodic arc	_	$2 \times 10^{-3} \text{ mm}^3/\text{g} @ 90^\circ$, 90 m/s, 50 μ m alumina	[24]

S. No.	Coating	Total Thickness	Deposition Method	Hardness	Erosion Rate	Ref.
3	Nanocomposite of nc-TiCN and a-SiCN having grain size of 5 nm to 7 nm	~ 20 to 35 µm	Plasma enhanced magnetron sputtering	20 to 40 GPa	$\sim 1 \times 10^{-3} \text{ mm}^3/\text{g}$ @ 90°, ASTM-G76-04, 50 μ m alumina. (Erosion rate one order lower than the TiN coating)	[113,142, 163]
4	Ti/TiAlN, Ni+Ti/TiN nanolayered multilayered coatings with two segments. Bi-layer thickness: 300 nm to 5 μm	~ 10 to 60 µm	Filtered arc deposition	20 to 80 GPa	Mass loss: 0.1 to 0.2 μg/g, 90°, 50 μm alumina. Test standard: GEET50TF121	[165]
5	Ti/TiN multilayer coating with stress absorbing metal layers. Bi-layer thickness~ 7.5 nm	~ 9 µm	Unbalanced magnetron sputtering	~ 17 GPa	$\sim 5.2 \times 10^{-3} \text{ mm}^3/\text{g}$ @ ASTM-G76-13, 90°, 100 m/s, Test temperature: 400 °C, 50 µm alumina	[132]
6	Ti/TiN multilayer coating with energy-absorbing nano-porous metal layers. Bi-layer thickness~ 7.5 nm	~ 9 µm	Unbalanced magnetron sputtering	~ 18 GPa	$\sim 2 \times 10^{-3} \text{ mm}^3/\text{g}$ @ ASTM-G76-13, 90°, 100 m/s, Test temperature: 400 °C, 50 μm alumina	[153]
7	CrAlTiN, TiN, Single phase coating	~ 16 µm	Arc physical vapor deposition	_	Deposited blades were tested in a test rig with a sand erodent at ~151 m/s speed. CrAlTiN coating showed better erosion resistance compared to TiN	[120]
8	Ti/TiN/TiAlN multilayer coatings	~10 µm	Cathodic arc system	_	~0.060 mg/g and 0.024 mg/g at the erosion angle of 90° and 30°, respectively	[166]
9	Ti _{0.52} Cr _{0.48} N (20 nm)/Ti _{0.40} Cr _{0.60} N (40 nm) multilayer coating	~20 µm	Cathodic arc system	28 GPa	~217 g/cm ² @ 70 m/s, 90°, silica particles of 200 μm size, ASTM G76.	[167]

Table 2. Cont.

4. Scope for Future Research

Solid particle erosion is a complex process that involves cutting, fatigue, cracking, and melting. Several studies were conducted to understand the erosion mechanism in ductile and brittle materials. In addition, different types of protective coatings were previously developed. However, there are still some gap areas that need the focus of the researchers to develop efficient erosion-resistant coatings.

- 1. Even though the erosion mechanism is well understood in ductile and brittle materials, the influence of temperature on the erosion process is still not fully established. It is an important aspect because gas turbines are usually operated at high temperatures. Different materials behave differently with increasing temperature in terms of corrosion and mechanical properties. So, more experimental and simulation studies on various aerospace materials are required to establish the role of temperature on the erosion process. These studies will help in choosing suitable materials to develop erosion-resistant coatings for high-temperature aerospace applications.
- 2. Most of the previously reported protective coatings for compressor blades were tested at RT. The erosion resistance performance of the developed protective coatings should

be tested at high temperatures so that the tested results match the real conditions of the gas turbine blades.

- 3. Third-generation nanocomposite and nanolayered multi-layered coatings with less than 10 nm grain size proved to perform better against the SPE. However, there is still limited literature available in this area. The development of erosion-resistant nanocoatings with different compositions and architectures can lead to the establishment of potential erosion-resistant coatings.
- 4. Recently, our group reported that the energy-absorbing nanoporous metal layers can be used in erosion-resistant coatings [153]. The coating performance against the SPE significantly improved after introducing the energy-absorbing nanoporous metal layers. More research in this direction should be carried out to improve the efficiency of SPE-resistant coatings.
- 5. Erosion-resistant coatings need higher thickness values (minimum thickness $\sim 7 \mu m$) with good mechanical properties. In order to achieve the required coating with low production costs, cost-effective precursor materials, higher growth rate techniques such as cathodic arc, and novel coating designs should be identified.

5. Conclusions

In this review, initial reports on the adverse effects of SPE on gas turbine engines and other sectors are discussed along with the fundamental concepts. However, the focus of the review has been more towards aerospace applications, and in particular, gas turbine engine components. Due to SPE, the efficiency of gas turbines decreases, and maintenance costs and oil consumption increase significantly. As the gas turbine engine compressors are operated at high temperatures, solid particle erosion studies conducted at high temperatures on different materials, particularly aerospace alloys such as Inconel, Ti6Al4V, SS, etc. are reviewed. Depending on the impinging angle and test temperatures, the erosion rate of aerospace alloy materials initially decreased in some cases and increased in others with increasing temperature. However, most of the studies showed an initial decrease in erosion rate with increasing temperature and then an increase at high temperatures. The decreased erosion rate was ascribed to the increased toughness, rate of dynamic recovery, and load bearing capacity of the materials. Materials with ionic and metallic bonds, which are flexible and non-directional bonds, contribute to increased toughness and rate of dynamic recovery with increasing temperature, thus leading to a reduced erosion rate with increasing temperature.

The influence of oxidation and corrosion on the erosion process was also discussed. Depending on the materials, oxidation does not influence erosion at initial temperatures (\leq 400 °C); however, at higher temperatures, oxidation plays a significant role in the erosion process. Whereas the erosion process in a corrosion environment causes accelerated erosion due to the erosion–corrosion combined effect.

The developed SPE-resistant coatings were classified into three generations. Single phase first-generation hard coatings, which failed to protect the substrates against normal impingement due to their low toughness. The introduction of metal/ceramic multilayered design (second-generation coatings) resulted in reduced stress and improved toughness. These coatings performed better than the single-phase nitride coatings against SPE. Further improvement in the coating efficiency against SPE was achieved by nanocomposite and nanolayered multilayered coatings (third-generation). Better erosion resistance of nanocoatings was ascribed to lower grain size (≤ 10 nm) and a high number of interfaces between metal and ceramic materials. In addition, energy absorbing nanoporous metal layers were used in a nanolayered Ti/TiN multi-layered coating to increase the erosion resistance performance even further. Coatings with a high number of interfaces, a lower grain size, a high load bearing capacity, lower internal stress, better adhesion, and a high thickness were found to work efficiently against SPE.

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