Degradation mechanism for high-temperature sliding wear in surface-modified In718 superalloy

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Abstract: This technical paper deals with high-temperature dry sliding wear behavior and its mechanism of Al2O3–50TiO2 (A50T) coating on Inconel 718 alloy. The sliding wear behavior of the A50T coating on Inconel 718 alloy was investigated using a pin on disc equipment at 500°C with varying parameters like normal load and sliding velocity. Scanning Electron Microscope (SEM) features of worn samples reveal that ploughing, deep grooving and splat exfoliation are the dominant wear mechanisms of A50T coating at 10 N, while at 30 N, they are crack extension, crack deflection, crack bridging and splat exfoliation.

Subjects: Manufacturing Technology; Corrosion-Materials Science; Materials Processing; Metals & Alloys; Surface Engineering-Materials Science

Keywords: superalloy; plasma spray; friction; sliding wear mechanism; A50T; SEM; 3D surface topography

1. Introduction
Wear can be defined as the interaction between surfaces resulting in the removal of material.

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PUBLIC INTEREST STATEMENT
Regarding mechanisms with which sliding wear phenomenon happens, virtually no report is available in literature on alumina–titania (Al2O3–50TiO2) on In718 superalloy at 500°C. Thus, the main objective of this technical paper is towards elucidation of different sliding wear mechanism of alumina–titania (Al2O3–50TiO2) at 500°C, with varying the parameter normal load. Further, it is well known that mechanism only always determines the wear rate and vice versa is not true. Hence, this paper lays more emphasis on mechanistic aspects only which provides depth information for readers.
Wear can occur in a number of modes, that include abrasive, adhesive and erosive wear which categorize the field of tribology (Sarkar, 1976). In particular, wear at elevated temperature is a serious problem in a large number of industrial applications such as power generation, high temperature bearing, gas turbine seals, nozzle, turbine blades in gas turbines, moving assemblies for hypersonic aircraft and missiles (Bulut Coskun, Aksoy, & Mahmut, 2012; Hamed & Tabakoff, 2006; Sandeep, Shukla, & Chona, 2011; Thirugnanasambantham & Natarajan, 2015).

Wear at elevated temperatures brings further complications in engineering materials, due to loss of mechanical strength of materials and alterations in the surface conditions leading to changes in adhesion between the surfaces caused by the joint action of temperature and tribological parameters (Du, Datta, Inman, Geurts, & Kubel, 2003). It degrades the performance and lifetime of mechanical components and results in economic loss. Thus, mechanical components operating in aerospace often require strength to withstand the synergetic attack combining wear and high temperature.

In recent times, gas turbine technology for power generation and for aeroengine applications places an increasing demand on the use of Ni-based superalloys (Adam Khan, Sundarrajan, & Natarajan, 2014; Gurrappa, Yashwanth, & Gogia, 2011; Restall & Stephenson, 1987; Tabakoff, 1999). Although the Ni-based superalloys possess adequate mechanical strength and creep resistance at the high operating temperatures, they often lack resistance to the wear environments. In such environments, alumina-based ceramic protective coatings are widely employed to improve the wear resistance of Ni-based superalloys. Alumina-based ceramic coatings are used in a wide range of applications including power generation equipment, aircraft engines, land-based and marine turbines and ships (Turunen et al., 2006).

Alumina-based ceramic coatings exhibit high hardness, high thermal resistance, friction and wear resistance at elevated temperatures (Benoy & Dellacorte, 1996; Trabelsi et al., 1989; Yust & Carignan, 1985). Although alumina-based ceramic coatings exhibit high hardness, its main drawback is its brittleness. The addition of titanium oxide to alumina leads to a balanced properties maintaining sufficient hardness and substantially increasing the coating toughness. Titanium oxide has a lower melting point and plays a role of binding alumina grains to achieve coatings with a higher density (Yılmaz, Kurt, Demir, & Tatlı, 2007). The frictional bond between the intersplats is favored by the alumina–titania glassy phase with a lower melting point that results in the good intersplat cohesion. Thus, alumina–titania coatings exhibit equilibrium of properties, maintaining enough hardness and coating toughness at elevated temperatures.

The wear mechanisms of the brittle material and hard coating are characterized by plastic deformation, intergranular cracking, Hertzian cone crack and median/radial crack system. Lawn, Evans, and Marshall (1980) reported the details on wear behaviors of the ceramic coating under dry sliding, revealing that the modes of deformation and fracture of ceramic coatings strongly depend on the normal load.

Regarding mechanisms with which sliding wear phenomenon happens, virtually no report is available in literature on plasma-sprayed Al₂O₃–50wt-%TiO₂ (A50T). Thus, the main objective of this work is to study the friction and wear mechanisms of plasma-sprayed Al₂O₃–50wt-%TiO₂ (A50T) coating by varying the normal load, under dry sliding condition using a pin on disc universal tribometer at 500°C.

In addition to the SEM, the characterization of 3D surface topography is essentially important in the field of tribology that provides useful information for analysis of wear scar surfaces (Demkin & Izmailov, 1991; Dong, Sullivan, & Stout, 1992; Sugimura & Kimura, 1984). In the present investigation, high-resolution optical profilometer–universal tribometer has been used for scanning the
surface of worn samples of A50T to obtain 3D surface topography images. The 3D surface topographic images provide a comprehensive picture of worn samples, in predicting the depth of worn surface and complete track of the wear scar.

2. Materials and methodology
In the present study, Inconel 718 superalloy is chosen as the substrate, and its nominal chemical composition (in wt%) is summarized in Table 1. Hardness of Inconel 718 is 40 Rockwell C Hardness (HRC). Samples with dimension of 8 mm × 5 mm sliced using wire electrical discharge machining (wire-cut) were coated with Al$_2$O$_3$–50TiO$_2$ using atmospheric plasma spray process with the coating parameters of industrial standards as mentioned in Tables 2 and 3. Dry sliding wear tests were carried out at 500°C as per ASTM G76 standard on specimens with 250–300-mm-thick Al$_2$O$_3$–50TiO$_2$ coating, using pin on disc tribometer, to simulate sliding wear of the coatings. Temperature is measured with inbuilt sensor in the sliding test equipment, provided by the maker M/s DUCOM, Bangalore, to conduct the sliding test as per ASTM G 76 standard.

The temperature was elevated by setting the furnace temperature in the tribometer, and it was measured by attaching the temperature sensor to the specimen holder. The testing conditions are summarized in Table 4.

Optical microscope (Leica, D2700M, GmbH) and scanning electronic microscope (TESCAN—VEGA 3, Czech Republic) were used to observe the surface morphology of the worn samples to investigate the sliding wear mechanisms. Optical profilometer is one of the main components in universal

| Table 1. Material chosen and their nominal chemical compositions (in wt%) |
|-----------------------------|---------------|---------------|---------------|---------------|---------------|---------------|---------------|---------------|---------------|---------------|
| Material | Ni | Fe | Cr | Mo | Ti | Al | Mn | Cu | C | S | Si | Cb + Ta |
| In 718 | 52.5 | 18.5 | 19 | 3.05 | 0.9 | 0.5 | 0.8 | 0.15 | 0.04 | 0.008 | 0.18 | 5.13 |

| Table 2. Parameters for atmospheric plasma spray coatings |
|-----------------------------|-----------------------------|-----------------------------|
| Spraying condition | Parameters | Units |
| Plasma arc current | 450–500 | A |
| Arc voltage | 60–70 | V |
| Plasma gas (argon) | 1.3–1.5 | ×10$^{-3}$ m$^3$ s$^{-1}$ |
| Secondary gas (hydrogen) | 0.3–0.4 | ×10$^{-3}$ m$^3$ s$^{-1}$ |
| Powder feed rate | 0.66–0.83 | g s$^{-1}$ |
| Torch to base distance | 76–125 | mm |

| Table 3. Details of powder used for coatings |
|-----------------------------|-----------------------------|-----------------------------|
| Chemical composition | Particle size, μm | Reference |
| Al$_2$O$_3$–50% TiO$_2$ (A50T) | 30 ± 5 | METCO131VF |

| Table 4. Sliding test parameters |
|-----------------------------|-----------------------------|
| Specimen size (mm) | 8 × 5 |
| Pin material | A50T |
| Disc material | Alumina |
| Sliding velocity (m/s) | 1.5, 2, 2.5 |
| Normal load (N) | 10, 20, 30 |
| Temperature | 500°C |
tribometer, which acts as surface profilometry tool to capture and analyze the surface topographical images of worn surface.

3. Results and discussion

Figures 1–3 illustrate the variations of the coefficient of friction (COF) of the samples with increasing distance at normal loads of 10, 20 and 30 N. The COF was recorded throughout each pin-on-disk test by utilizing the tangential and normal load sensors of the tribometer. When two contacting surfaces slide against each other, a frictional force is generated opposite to the direction of sliding. During sliding, the frictional force is considered to be exerted in a direction perpendicular to the normal load. The ratio of these forces is equal to the COF.

Figures 1–3 show that the friction coefficient of the A50T specimens under dry sliding condition increases with increasing normal load. As the applied load increases, the number of asperity interactions between the rubbing surfaces increases, resulting in increased friction coefficient. The gradual increase of the friction coefficient can be associated to the real contact area between the rubbing surfaces. At the beginning of sliding, the asperities of the surface limit the sliding speed. Meanwhile, the friction leads to removal of some of the asperities. This removal of the asperities would change the applied stress on the worn

![Figure 1. Variation of coefficient of friction with distance at 1.5 m/s.](image1)

![Figure 2. Variation of coefficient of friction with distance at 2 m/s.](image2)
surface, which could change the mechanism from two-body to three-body abrasion involved in the wear process.

The wear mechanism transition causes sudden changes in the tangential (or friction) force, resulting in the friction coefficient fluctuations in a wavy form.

Figures 1 and 2 show the oscillations in the friction coefficient behavior of the A50T specimens; this should be attributed to the interactions between the surface asperities and the induced dynamics of the contacting bodies and entrapment of wear particles. Some wear particles may favor rolling processes, which lower the friction coefficient. Figure 3 illustrates that the friction coefficient of the A50T specimens for all different loads that initially increased with increasing sliding distance until a peak value was reached, and then, it gradually approached steady state. This is the so-called running-in stage. The initial running-in corresponds to the contact of the asperities between the rubbing surfaces.

A50T coatings exhibit balanced properties of hardness and toughness at elevated temperatures. These material properties suppress the junction growth at asperity contacts in friction and the real area of contact. Hence, the friction coefficient of A50T coating does not exceed to a value above 1.0, as shown in Figures 1-3. It is well known that the lowest COF results in higher wear resistance. Investigating the wear mechanisms is more important than the knowledge of actual wear rates from the design point of view (Eyre, 1978). Generally, normal load plays a predominant role in determining the sliding wear mechanisms (Stott and Jordan, 2001).

3.1. Wear mechanisms of A50T-coated surface at 10 N
When two sliding surfaces come into contact, normal and tangential forces are transmitted through contacting asperities. Figure 4a shows formation of ploughing on the surface of the coating due to tangential sliding action of asperities. Tangential sliding action of asperities causes local pressure between contacting asperities; as a result, Hertzian contact stresses arise at the interface of contacting asperities, resulting in ploughing. The role of ploughing mechanism is to reduce the energy of sliding asperity during the sliding action; therefore, much less kinetic energy is available for initiating and propagating the cracks along the grain boundary of A50T coating. Further repeated sliding action induces residual stresses in the coating. To attain the equilibrium configuration, residual stress inside the plastic zone is released in the form of lateral cracks along the sides of the grooves. At this stage, there will be transition of mechanism from ploughing to crack initiation and propagation along the splat boundary of the coating.
and finally resulting in splat exfoliation (Figure 4b), as corroborated by the previous researcher (Braza et al., 1989).

After splat exfoliation, the next sublayer becomes as a new sliding surface. This process is continued, resulting in crater. Attributing to the insufficient load carrying capacity due to continuous action sliding cycles, the removed splat is crushed into smaller hard particles and remain in the contact zone between pin and disc and promote wear damage to the coated surface. Figure 4c reveals the entrapment of hard particles in the crater, causes high abrasion stress which could change the wear mechanism from two-body to three-body abrasion. Wear transition of A50T coating causes significant change in tribocontact surface topographical features such as peaks and valleys, as shown in Figure 9a. The presence of third bodies remaining in the contact zone leads to an increase in the applied stress on the worn surface, causing more localized fracture as shown in Figure 4d.

### 3.2. Wear mechanisms of A50T-coated surface at 30 N

The lateral cracks are mainly responsible for sliding wear in ceramic-based coatings. The magnitude of extension of lateral crack and propagation depends on the normal load. As the applied load increases, the number of asperity interactions between the rubbing surfaces increases, resulting in increased friction coefficient. At higher loads, the intensity of residual stress will increase due to the higher friction coefficient on the interface, under the action of normal and tangential load. Residual stresses within the plastic zone are released in the form of lateral cracks. Lateral cracks are initiated from sites of preexisting surface flaws, and subsequently, cracks extend (Figure 5a) under the normal and frictional stresses.

Presence of TiO₂ in alumina plays a vital role in binding the alumina grains effectively to achieve good frictional bond between the intersplats, which provides shielding mechanisms to the crack tip through crack deflection, crack branching and crack bridging. Crack extends with local deflections from its general crack path direction. Figure 5b reveals that series of crack deflections and crack branching...
results in causing the R-curve behavior in A50T. The concept of R-curve behavior is that fracture resistance increases with crack extension, as corroborated by the previous researchers (Cho, Hockey, Lawn, & Bennison, 1989; Srinivasan & Scattergood, 1991; Steinbrech, Reich, & Schoarwachter, 1990).

Figure 5. SEM images showing sequence of sliding damage of A50T surface at 30 N, 2 m/s. (a) Crack extension, (b) crack deflection, (c) crack bridging and (d) splat exfoliation.

Figure 6. SEM image of worn surface of A50T at 30 N and V = 2 m/s. (a) Visible of sublayer splats, (b) sublayer splat exfoliation and (c) deep crater.
R-curve behavior in A50T increases the fracture resistance, due to stress transfer or kinetic energy transfer behind the crack tip. This stress transfer occurs through crack extension, crack deflections and crack branching mechanisms, which reduces the intensity of crack tip for further propagation. Therefore, R-curve behavior influences the energy consumption during crack propagation. Hence, additional energy is required for the crack tip to propagate the crack. Further, subsequent sliding action on surface causes crack propagation along splat boundary, resulting in formation of crack bridging, shown in Figure 5c. Crack bridging promotes the frictional bond between the intersplats, which resist the splat exfoliation. Therefore, additional energy is required to overcome the bridging
friction and to separate the crack surfaces, which increases the fracture resistance of the coating (Erdogan & Joseph, 1989). Due to further sliding action, crack bridges get damaged and finally resulting in splat exfoliation, as shown in Figure 5d. Figure 6a reveals subsurface splat due to the subsequent removal of material. On the worn surfaces of the coating, the existence of sublayer splat exfoliation (Figure 6b) gives considerable evidence that fracture occurs along the sub-splat boundary. Figure 6c shows the deeper crater in the coated surface due to the excessive sublayer splat exfoliation material removal during the cyclic sliding action on worn surface.

Due to the continuous action sliding cycles, the removed splat is broken into smaller hard particles and entrapped in the deeper crater. The entrapment of hard particles in crater (Figure 7a) leads to an increase in the applied stress on the worn surface and induces the dynamics of the contacting bodies, which could change the mechanism from two-body to three-body abrasion involved in the wear process, resulting in sudden changes in the tangential (or friction) force which eventually causes the friction coefficients to fluctuate in a wavy form (Figure 2) and also causes significant

Figure 9. (a) 3D surface topography of the worn surface at 10 N and V = 2m/s. (b) Topography of the crater depth of worn surface at 10 N and V = 2m/s. (c) Wear track profile of worn surface at 10 N and V = 2m/s.
change in a tribocontact surface topographical features such as central depression and formation of well-defined peaks and valleys, as shown in Figure 8a. Wear mechanism transition of A50T coating increases the contact stresses at their interface, causes rapid grain chipping from the surface of the coated material and finally results in severe damage, as shown in Figure 7b.

Generally, the magnitude of depth and size of crater are proportional to the normal load. The depth of wear scar is significantly larger at 30 N (Figure 8a-c) than at 10 N (Figure 9a-c). This is mainly due to crack propagation and splot exfoliation during sliding action, causing more effective damage to the target surface.

4. Conclusion

Wear mechanisms of the A50T coating is greatly influenced by the normal load.

- At 10 N, the mechanism of sliding wear of A50T coating on In718 occurs through ploughing, crack initiation and propagation along the splot boundary and splot exfoliation.
- At 30 N, the sliding wear mechanism of A50T is found to be through crack extension, crack deflection, crack bridging, splot exfoliation, sublayer splot exfoliation and deep crater
- At elevated temperature, the presence of TiO$_2$ in alumina effectively binds the alumina grains which imparts the R-curve behavior through crack deflection, crack branching and crack bridging.
- The synergic effect of crack deflection, crack branching and crack bridging reduces the effectiveness of energy transfer to the target material, resulting in increasing the fracture resistance of coating, thereby enhancing the wear resistance.

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