Introduction

Currently, the implantation of an orthopedic joint prosthesis is a very common procedure that has proven to be an effective treatment to enhance mobility and reduce pain, but is still not optimized for longevity. In 2006, approximately 10% of all total knee and hip replacements failed and required renewal (AAOS, 2006). The most significant cause for these orthopedic failures is the formation of particulate debris, which is a direct result of wear occurring at the interface of the bearing surfaces of the implant. Ions in the blood plasma also initiate corrosion of the metallic surfaces which weakens the implant, in addition to forming mechanistically similar particulate matter. This particulate matter formation initiates failure through induction of osteolysis, the disappearance of surrounding bone. Orthopedic implant materials are also subject to ion out-diffusion, where potentially hazardous ions leak from the implant metal, causing carcinogenic reactions, allergenic reactions, or contact dermatitis (IARC, 1990).

Typical modern orthopedic implants are fabricated from material with a very high elastic modulus, in order to minimize compliance when weight-bearing, but this mismatch with the elastic modulus of bone causes a resorption of the bone surrounding the implant due to stress shielding which fails to physiologically load the bone. The most common material combinations that are used in the bearing surfaces of orthopedic implants today are metal-on-polyethylene, ceramic-on-polyethylene, ceramic-on-ceramic, and metal-on-metal. Polyethylene containing implants suffer from the greatest wear, estimated at 100 to 200 µm each year, and are likely to initiate osteolysis. Alternatively, implants consisting of bulk, wear-resistant ceramics, such as Al₂O₃, typically coincide with very high elastic modulus values. As discussed above, this mismatch in elastic modulus causes a reduction
in supporting bone density over time, culminating in implant failure. Metal-on-metal implants contain the most potential for corrosion and ion out-diffusion, leading to potential adverse reactions in the body in addition to the occurrence of osteolysis. The ideal implant material combination would therefore be highly wear-resistant at load bearing surfaces, have high resistance to corrosion of metal components, minimal ion out-diffusion, while also possessing an elastic modulus close to that of bone. Atomic layer deposition (ALD) is a common method applied for the purposes of high-κ dielectrics, and it has become widely accepted that ALD films of Al₂O₃ exhibit high conformity, precise thickness, and pinhole free deposition. These characteristics indicate ALD as a method with strong potential for the resolution of existing complications in modern orthopedic implants. This thesis examines the enhancement and effects that ALD thin films have at the interface between the substrate through use of nanomechanical testing.
Background

Sample Synthesis Techniques

Many materials have the propensity for the formation of a native oxide layer on their surface due to reaction with ambient air. This native oxide layer is extremely thin, with a typical thickness of approximately 15 Å (Stewart, 2003). Reactive ion etching (RIE) is a method that alters the surface characteristics of a material through etching that occurs through ion bombardment and/or chemical reactions at the surface. The gas of interest is introduced to the chamber at very low pressures on the order of 0.1 Pa. Power is input to the system to excite the gas to a plasma that contains ions of the original input gas. These ions are then accelerated toward the sample surface through a voltage difference that is created between the top and the bottom of the RIE chamber (Jaeger, 2002). This study specifically applies RIE that uses oxygen or argon gas as the feed gas. The bombardment of the native oxide layer with oxygen ions results in a combination of further oxidation and removal of the native oxide, whereas inert argon ions only remove the native oxide (Jaeger, 2002).

ALD is a method used in industry for the manufacturing of high-κ materials due to the extreme precision that the method allows. ALD films are typically very thin, ranging from 10 nm to no more than 1000 nm. The reactions that occur in ALD are contained in a sample chamber at high temperature and very low pressures. Alternating precursors are then pulsed into the reaction chamber, and immediately thereafter purged to allow for layer by layer deposition on to the sample surface. The Al₂O₃ coating is created through the pulsing of Al(CH₃)₃ (TMA), which bonds with surface hydroxyl groups through loss
of a methyl group. After purging excess TMA, water is pulsed to substitute the remaining methyl groups with oxygen bridges and hydroxyl groups. This sequential pulsing of TMA and water is illustrated in Figure 1.

![Figure 1: Schematic of an ALD reaction cycle. In the left portion of the figure, TMA is pulsed into the reaction chamber and reacts with the substrate native oxide, forming a methane byproduct. Residual TMA and byproduct methane is subsequently purged. Water vapor is then pulsed and reacts with the remaining methyl groups to form oxygen bridges and new surface hydroxyl groups. Purge follows, and the cycle is complete. The rightmost portion of the figure illustrates the multi-layer structure of Al₂O₃ (Bryan Christie Design, 2008).](image)

**Nanomechanical Testing**

After sample preparation, nanomechanical testing was performed to allow for further understanding of how the Al₂O₃ film behaves at the surface of the metal substrate. A UBI-1 nanomechanical testing system was used, and contains several components, including a granite base, a xyz staging system, an optical camera system, a piezo scanner, a transducer, a vibration isolation system, an acoustic enclosure, and the necessary
electronics for data acquisition and processing. The piezo scanner, named the triboscanner, provides fine scale positioning of the tip that probes the sample. This operates on the use of piezoelectric ceramics, which bend in response to when voltage is applied across them. The cylindrical piezoelectric ceramic tube then will bend in the x or y direction, as controlled by a designated voltage input. The most important constituent is the three-plate capacitive force-displacement transducer, which coordinates output forces with a controlled input voltage (Cooper, 2007). The transducer operates through the use of two parallel electrodes with an electric field that is at its maximum at either electrode, while also being equivalent to zero at the area equidistant from each electrode. Each electrode has its electric field set up through an AC signal that is 180 degrees out of phase with the electric field of the other electrode. If a large DC offset is applied to the top plate, an electrostatic attraction will arise between the top plate and the center plate, which has the tip attached to it. This electrostatic attraction then causes the tip to impart a very precise, accurate force (Hysitron).

The UBI-1 nanomechanical testing is capable of performing indentations, wearing of designated areas, and scratching. In nanoindentation, a load versus displacement curve is created through the application of force that increases at a constant rate until a designated peak load has been reached. Afterward, the force is decreased at the same constant rate. This creates a load versus displacement curve similar to that displayed in Figure 2.
Figure 2. A sample plot of a load versus displacement curve for an indentation on fused quartz. The unloading slope is a critical parameter that the reduced modulus is directly dependent upon (Vieregge, 2010).

From the load versus displacement curve, reduced modulus \( E_r \), elastic modulus \( E \) and the hardness \( H \) of the material may be extrapolated. The reduced and elastic modulus characterize a material’s stiffness, whereas the hardness characterizes a material’s resistance to localized deformation (Callister, 2007). The reduced modulus is defined by

\[
E_r = \frac{dP}{dh \sqrt{\pi}}
\]  

(1)

where \( A \) is the contact area, \( P \) the applied load, and \( h \) the depth of penetration (Vieregge, 2010). The derivative of \( P \) with respect to \( H \) is indicated on Figure 2 as the slope of the unloading curve. The elastic modulus may be calculated from the reduced modulus as follows

\[
\frac{1}{E_r} = \frac{(1-v_i^2)}{E_i} + \frac{(1-v_s^2)}{E_s}
\]  

(2)

and the subscripts \( i \) and \( s \) correspond to the indenter material and the substrate material, respectively, while \( v \) represents the Poisson’s ratio (Vieregge, 2010).
Lastly, hardness is defined by the function

\[ H = \frac{P_{\text{max}}}{A} \]  

(3)

in which \( P_{\text{max}} \) describes the maximum applied load (Vieregge, 2010).

In scratch testing, normal force, normal displacement, lateral force, and lateral displacement must be measured, therefore requiring a transducer that acts in 2 dimensions. In a scratch test, a normal force is applied to the sample at a constant rate, while also moving towards a designated lateral position at a constant rate. Scratch testing yields important information regarding the interfacial adhesion of a thin film, such as \( \text{Al}_2\text{O}_3 \), to the underlying substrate. The critical load that initiates delamination of the film from the sample may be measured through observation of interfacial failure in the output data. Interfacial failure is marked by a combination of sudden changes in the slope of the lateral force and normal displacement as a function of time (Vieregge, 2010). This is caused by a sudden release of energy that occurs with the delamination of the thin film from the substrate (Vieregge, 2010). Understanding the critical load is very important, as higher critical loads typically correlate to better interfacial adhesion of the thin film to the substrate.

The UBI-1 nanomechanical testing system also has the capacity for scanning probe microscopy (SPM). It may form the image of a surface through raster scanning a probe across the sample surface at a relatively small normal force. Changes in normal displacement are correlated to changes in color darkness to form a gradient image of the surface. Nanowear testing modifies this capability to allow for the analysis of a sample’s resistance to wear. The probe initially uses SPM to scan over an untested surface and form an image. Subsequently, the probe imparts wear within the initially scanned surface
by scanning the surface at a higher set point normal force. In addition to this, repeated passes are made over the same surface. Afterward a second SPM image is taken over the same area as the first image, allowing for wear depth to be calculated from the gradients of the SPM images both before and after wear was imparted.
Materials and Methods

Sample Preparation

Sample preparation entailed substrate cleaning, followed by deposition via ALD of 200, 600 and 1000 nm $\text{Al}_2\text{O}_3$. Select samples also received RIE treatment prior to deposition. Polished 305 stainless steel (SS) sheets, cut into 15.5 mm disks were used as the substrate for all samples. All substrates were rinsed in acetone, then sonicated for 5 minutes in acetone. This same process of rinsing followed by sonication was then repeated using isopropanol alcohol then deionized water. Select sample groups also received RIE treatment of either oxygen or argon plasma prior to deposition. The same process parameters including exposure time, chamber pressure, flow rate, and power were used for etching in either plasma. The RIE chamber was first purged for 5 minutes with the selected gas at a flow rate of 25 sccm for 5 minutes at a pressure ranging between 0.4 to 0.7 Pa. Following purge, the selected gas was excited to a plasma using 200 W of power and allowed to flow through the chamber at the same flow rate and pressure as in purge. After RIE is performed, samples were removed and immediately stored in a nitrogen box. $\text{Al}_2\text{O}_3$ was then deposited through use of ALD. Prior to deposition, the ALD was prepared by preheating the chamber to 300°C, dropping the pressure to 12 hPa, and flushing the chamber with 600 sccm nitrogen gas through a controlled fluctuation in pressure. Once the chamber reached the set point temperature of 300 °C and pressure of 12 hPa, 15 minutes were allowed to fully stabilize and reach complete uniformity. The samples to receive deposition were then taken from the nitrogen box and placed within the chamber of the ALD in addition to a single silicon sample that serves as a watch. For
each cycle, TMA was pulsed for 0.1 seconds, and purged from the chamber for 2 seconds using N₂. H₂O vapor was then pulsed for 0.1 seconds, and subsequently followed by a 2 second N₂ purge. Each cycle was correlated with a specific thickness of 0.089 nm, and this was used to calculate the required number of cycles in order to achieve the target thicknesses of 200, 600 or 1000 nm. After each deposition, the film thickness of the silicon watch was measured through use of an ellipsometer to verify that the film thickness meets specification. Five points on each sample were measured for thickness to assess uniformity and to verify that thickness was within the allowable range.

In summary of sample preparation, three RIE treatments utilize i) oxygen plasma, ii) argon plasma, or iii) control (no plasma). Al₂O₃ film thicknesses deposited using ALD were of 200, 600, or 1000 nm. Therefore, two methods with three different treatments each give rise to 9 total samples. This is summarized below in **Table 1**.

**Table 1**: Summary of RIE treatments and deposition thicknesses of each sample fabricated and tested.

<table>
<thead>
<tr>
<th>Argon Plasma &amp; 200 nm Al₂O₃</th>
<th>Oxygen Plasma &amp; 200 nm Al₂O₃</th>
<th>200 nm Al₂O₃ only</th>
</tr>
</thead>
<tbody>
<tr>
<td>Argon Plasma &amp; 600 nm Al₂O₃</td>
<td>Oxygen Plasma &amp; 600 nm Al₂O₃</td>
<td>600 nm Al₂O₃ only</td>
</tr>
<tr>
<td>Argon Plasma &amp; 1000 nm Al₂O₃</td>
<td>Oxygen Plasma &amp; 1000 nm Al₂O₃</td>
<td>1000 nm Al₂O₃ only</td>
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**Nanomechanical Testing**

Nanomechanical testing, including scratch, wear, and nanoindentation, was performed on the fabricated samples using an UBI-1 nanomechanical testing system. For the nanoindentation tests, a three-sided, diamond, pyramidal, Berkovich tip was used in conjunction with a 1-D transducer. Therefore, in application to **Equation 2**, vᵢ and Eᵢ are
0.07 and 1140 GPa, respectively (Vieregge, 2010). Prior to any testing, the nanoindenter was calibrated with an air scratch that fixed the transducer constants, an H-pattern that calibrated positioning of the probe on the sample, and a control indentation test on fused quartz which verified output values for $E_r$ and H as aligning with standard values. Each sample received a grid of 25 different nanoindentations where the force of the first indent was 2000 µN. Each subsequent indent applied a constant interval of force greater than the one before it, until a force of 9000 µN was attained for the final indent. This variation in force was necessitated in order to see how the material properties change as a function of depth into the film and substrate. Each nanoindentation consisted of a 10-second loading to peak force ramp, followed by a 10-second hold at peak force, and then finishing with a 10-second unloading to zero force ramp.

Wear testing was performed as described prior through use of a 2-D transducer in conjunction with a diamond conical tip. Only the aforementioned air calibration was necessary for wear testing, as manipulation of output data was not as strict as in nanoindentation. In each wear test, an initial “pre” 8 µm by 8 µm SPM image was taken on unworn sample surface. In the center of the area where the “pre” SPM image was taken, wear was imparted on a 4 µm by 4 µm area. Each sample was worn with forces of 120, 200, 250, or 350 µN, with each force passing over the wear area ten times. This range of forces allows for comparison of how different samples respond to different magnitudes in wear force, i.e. if differences in wear resistance are more prominent at larger forces. After the wear area had been completed, a secondary “post” SPM image was taken over the same area as the “pre” image for visual and quantitative comparison.
Scratch testing utilizes a 2-D transducer and a conical diamond tip. Because scratch testing is the only test that acts in two dimensions, both an air calibration test and a scratch indentation axis test were necessary to calibrate the constants on the transducer in both dimensions. Similar to wear testing, “pre” and “post” SPM images were performed on the areas in which the scratch is made, as this allows for visual observation of changes in film adhesion. Ramped force scratch tests were performed on each substrate at designated peak forces that i) penetrate the film or ii) demonstrate delamination of the film. Peak normal forces of 3 µN and 5 µN were used for all 200 nm Al₂O₃ samples. The 3 µN force wears partially into the film, and the 5 µN force penetrates the film. Peak forces of 8 µN and 12 µN were used for all 600 nm and 1000 nm Al₂O₃ samples. In both the 600 nm and 1000 nm Al₂O₃ samples, the force of 8 µN demonstrates no delamination, whereas the force of 12 µN causes delamination of the film. In each test, the probe was moved laterally 8 µm along the surface at a rate of 0.5 µm/sec, and during this time, the normal force was ramped to the designated peak force.
Results and Discussion

(a) Elastic Modulus (GPa) vs. Depth into 200 nm Al2O3 Sample (nm)

(b) Hardness (GPa) vs. Depth into 200 nm Al2O3 Sample (nm)
(c) Elastic Modulus (GPa) vs. Depth into 600 nm Al2O3 Sample (nm)

(d) Hardness (GPa) vs. Depth into 600 nm Al2O3 Coating (nm)

- Argon Plasma and Coating
- Oxygen Plasma and Coating
- Coating Only
Figure 3a-f. 4a, 4c, 4e describe the elastic modulus as a function of depth into the 200 nm, 600 nm, and 1000 nm sample, respectively. 4b, 4d, 4f describe the hardness as a function of depth into the 200 nm, 600 nm, and 1000 nm sample, respectively. Expected results are confirmed, as all above figures show a transition from ALD Al₂O₃ properties to that of bulk 305 SS. No statistically significant difference is shown between samples of different RIE treatments.
Nanoindentation was performed on each of the samples cited in Table 1, and yields fundamental information regarding the mechanical properties of the Al$_2$O$_3$-305 SS junction. Knowledge of these properties serves to confirm conceptual expectations and justify response behavior, such as that observed in wear testing. Figure 3 presents a thorough detailing of the changes in mechanical behavior for different depths into the sample, different RIE treatments, and different thicknesses of Al$_2$O$_3$. Literature values report ALD Al$_2$O$_3$ on a silicon substrate as having a modulus of 220 ± 40 GPa and a hardness of 10.5 ± 1 GPa (Tapily, 2008). In the thinner 200 nm sample, corresponding to Figures 3a and 3b, it is observed that the elastic modulus is nearly identical to that of the bulk 305 SS, while the hardness is slightly higher than that of bulk 305 SS. Because of the significantly higher reported modulus and hardness of ALD Al$_2$O$_3$, the hardness may be slightly higher in the 200 nm coated samples because of the earlier emergence of effects of the deposited Al$_2$O$_3$ on hardness than on elastic modulus. Figures 3b and 3c correspond to indentations performed on the 600 nm sample, and illustrate significantly more influence by the ALD Al$_2$O$_3$ coating than in the 200 nm sample indentations. This is observed in the higher elastic modulus and hardness for all indentation depths than that of the 200 nm sample. Figures 3a, 3b, 3c, and 3d all demonstrate a decline in hardness and elastic modulus with increasing penetration depth, likely due to lessening of the influence of the ALD Al$_2$O$_3$ film. However, Figures 3e and 3f, corresponding to the 1000 nm Al$_2$O$_3$ film, demonstrate a relatively constant elastic modulus and hardness, where ALD Al$_2$O$_3$ properties exist for all indentation depths. It is also observed in Figures 3e and 3f that the elastic modulus and hardness for the depths tested are close to those reported in the nanoindentation studies of ALD Al$_2$O$_3$ on a silicon substrate. Values
for elastic modulus are slightly lower than those reported by the study, whereas values for hardness are slightly higher than those. The lower values for elastic modulus are expected, as silicon has a much greater elastic modulus of approximately 190 GPa, than that of bulk 305 SS, which has an average modulus of 130 ± 4 GPa (Hopcroft, 2010). Slightly larger hardness values may be attributed to dulling of the indenter tip over time. This creates discrepancies in the tip area, which is indirectly proportional to hardness, thereby creating potential for a systematic increase relative to literature reported values.

**Figure 3** also demonstrates that nanoindentation yields no consistent, observable difference in the hardness or elastic modulus in different RIE treatments of argon or oxygen from the untreated samples that have just the Al₂O₃ coating. This is due to the relative thickness of the films with respect to the native oxide. Variation in the native oxide will not be exacerbated when a normal force is applied because there are no direct strains at the interface between the oxide and the film.

Scratch testing examined the tendency of a film to delaminate from the surface, and it was observed that ALD Al₂O₃ exhibits very strong adhesion to the surface of the substrate 305 SS, which is likely attributed to the conformity of films deposited via ALD. The 200 nm Al₂O₃ films do not delaminate even when scratch depth into the sample extends to an average of 175 ± 0.1% of the film thickness. The 600 nm and 1000 nm Al₂O₃ films delaminate at an average of 102 ± 20% and 42.6 ± 3.8% of their respective film thicknesses. The delamination of these thicker Al₂O₃ films is shown below in **Figure 4** as AFM images that visually demonstrate the possible tip flexion that occurs upon delamination. The increased propensity of thicker Al₂O₃ films towards delamination from the surface of the substrate is possibly a result of thicker films manifesting more
independent material properties from the substrate material. It may be conceived that thicker Al$_2$O$_3$ films receive less interaction with the interface between 305 SS, and thereby act more separately, therefore increasing their likelihood for delamination. This concept is also reinforced by Figure 3, which illustrates the 1000 nm film as having properties unchanging with penetration depth and closely matching that of ALD Al$_2$O$_3$ only.

Figure 4. AFM forward topography image of (a) a 12mN nanoscratch on a 1000 nm Al$_2$O$_3$ sample and (b) a 8 mN nanoscratch on a 1000 nm Al$_2$O$_3$ sample, both with no etching treatment. Delamination is pictured in (a), whereas (b) depicts scratch without delamination. The cause for this difference in pattern is still undetermined, but may be attributed to tip flexion.
Figure 5 indicates that argon and oxygen plasma etching treatment bears a possible effect on how the sample material responds to a ramped scratch of 8 and 12 mN in normal force. Although there are no differences between the oxygen plasma or argon plasma treatments themselves, it is shown that the sample with the Al2O3 delaminates at a depth approximately 20% greater than that of the treated samples. This is inconsistent with the observation of closely matching mechanical properties in RIE treated samples with the Al2O3 coating only samples. The difference in scratch resistance of the Al2O3 coating only samples from the RIE treated samples may be conceptually validated because scratch testing acts in both the normal and lateral directions, whereas nanoindentation only occurs in the normal direction. This would suggest that the native oxide provides more scratch resistance than do samples with a native oxide that has been diminished as a result of RIE treatment. In delamination, the most strain is applied at the interface.
between 305 SS, the native oxide, and the Al₂O₃ film, as this is where delamination occurs. 305 SS, even polished, has a large quantity of surface irregularities, which may be exaggerated by a thicker native oxide. A more irregular surface would thereby provide more surface area for the binding of Al₂O₃. More binding of Al₂O₃ at the interface would therefore produce a tighter bond at the interface with the native oxide, providing greater resistance to delamination.

Figure 6. Scratch depth at specified peak load and as a function of film thickness for the Al₂O₃ coating only samples. Material response to ramped normal forces show that thicker films receive less penetration with respect to thinner films, and that Al₂O₃ coating provides more scratch resistance when compared to the bulk 305 SS.

Figure 6 illustrates how the sample responds to scratches of 8 mN and 12 mN. This figure is consistent with the observation made in Figure 3, where thicker films less readily undergo penetration along the normal direction than do thinner films. Al₂O₃ films
in general also exhibit larger resistance to penetration over the bulk 305 SS alone, indicating that ALD Al$_2$O$_3$ promotes scratch resistance.

![Figure 7](image)

**Figure 7.** Wear depth in the normal direction, as measured by SPM image gradient, as a function of surface treatment, sample thickness, and wear force. Wear depth generally conforms to the expected trend of increasing wear depth with increasing force. Al$_2$O$_3$ films are also shown to provide substantial wear resistance with respect to the bulk 305 SS.

Wear testing yields results indicating how a coating changes a material’s resistance to wear in response to a constant normal force shearing across the sample surface. **Figure 7** demonstrates that ALD Al$_2$O$_3$ provides substantial protection of the underlying substrate material by reducing wear by up to 74.1 ± 5% of the resulting values for wear depth of the bulk 305 SS samples. There is a possible trend where, generally, greater forces impart larger wear depths than smaller forces, thereby coinciding with expected behavior. There are no observable or statistically significant differences between the oxygen or argon RIE treatments, and there is also no distinguishable difference of these RIE treated samples from samples with only Al$_2$O$_3$. This reinforces with the observation in nanoindentation,
which suggested that RIE treatment plays an insignificant role in the behavior of ALD
Al$_2$O$_3$ along the normal axis. **Figure 7** also yields a possible result where all samples
receive approximately the same wear depth for a given wear force, regardless of the
thickness of the Al$_2$O$_3$ film. This does not align with the previously observed behavior
where scratch resistance increased with larger film thicknesses. One reason for the
deviation of wear resistance from scratch resistance may be in the method through which
wear depth is measured. In scratch testing, scratch depth is measured through the normal
axis of the 2-D transducer, whereas wear depth is measured through an estimate from the
gradients in color that exist on the SPM image. As wear depth progressively gets deeper,
the gradients in color are unable to fully capture depth through color difference, therefore
providing a large source of variation.
Conclusion

This study has explored the application of ALD Al$_2$O$_3$ for the potential improvement of modern day orthopedic implants. Nanoindentation was performed for the examination of the mechanical properties of ALD Al$_2$O$_3$ at the interface with 305 SS. Mechanical properties follow expected trends where the elastic modulus and hardness of thinner films of Al$_2$O$_3$ align relatively closely with that of bulk 305 SS. Both the 200 nm and 600 nm Al$_2$O$_3$ samples demonstrate a trend of hardness and modulus transitioning in mechanical properties from those of the ALD Al$_2$O$_3$ film to those of the bulk 305 SS with increasing penetration depth. The 1000 nm sample receives relatively minimal influence from the underlying substrate, remaining roughly constant at values that closely align with literature values for ALD Al$_2$O$_3$. No observable differences exist between the different RIE treatments, or from the untreated samples because no direct strains are imparted at the interface between the native oxide.

Scratch testing demonstrated that ALD Al$_2$O$_3$ films exhibit excellent adhesion that increases with lower film thicknesses. The 200 nm Al$_2$O$_3$ films do not delaminate even when scratch depth into the sample extends to an average of 175 ± 0.1% of the film thickness. The 600 nm and 1000 nm Al$_2$O$_3$ films delaminate at an average of 102 ± 20% and 42.6 ± 3.8% of their respective film thicknesses. It is also observed that thicker films possess greater scratch resistance over thinner films. Although there are no differences between the two RIE treatments, it is shown that there is a possibility that samples with a diminished native oxide as a result of RIE are less scratch resistant than samples with no RIE treatment and only an Al$_2$O$_3$ coating. This may be because a thicker native oxide
provides a larger surface area for binding of Al₂O₃, and that strain is imparted directly at this interface.

Wear testing revealed excellent protection of the substrate material, reducing wear by up to 74.1 ± 5% of the resulting values for wear depth of the bulk 305 SS samples. Wear depth generally increases with the force imparted on the sample surface, thereby obeying expected results. No distinguishable differences exist between the RIE treated samples, or from the untreated samples. Similarly, with respect to the control, wear resistance is observed to be relatively constant regardless of film thickness. This may be attributed to the magnitude of difference in wear resistance between the coated samples and bulk 305 SS.
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