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HIGH-TEMPERATURE TRIBOLOGICAL CHARACTERISTICS OF SILVER AND GOLD COATINGS ON CERAMICS PREPARED BY ION-BEAM-ASSISTED DEPOSITION*

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ABSTRACT

An ion-beam-assisted deposition (IBAD) system was used to deposit silver and gold coatings on polycrystalline α-alumina (Al_2O_3) substrates for tribological studies at temperatures to 400°C. The wear tests were performed with an oscillating ball-on-flat type of test apparatus as a partial simulation of ring/liner motion and contact geometry in actual engine systems. The test results showed that without a surface coating, both the wear rates and the friction coefficients of Al_2O_3/Al_2O_3 test pairs were quite high, and increased substantially with temperature. In contrast, the wear of flats coated with silver and gold was at unmeasurable levels, even after sliding tests of 110,000 passes. The wear of balls (uncoated) sliding against the Ag- and Au-coated flats was reduced by factors of 45 to more than 500 depending on coating type and ambient temperature. The friction coefficients of pairs with an IBAD-Ag or Au coating were in the range of 0.32-0.5.

INTRODUCTION

Tribomaterials for use in advanced energy conversion and utilization devices (e.g., low-heat-rejection diesel engines, ball- and roller-bearing systems, advanced turbomachines) are required to operate under severe tribological and environmental conditions (e.g., high service temperatures, aggressive environments, high speeds, and extreme contact pressures). These operating conditions exceed the capabilities of most current lubricants and tribomaterials. Fine ceramics, owing to their excellent strength, hardness, creep resistance, and good insulating properties, appear very promising for these stringent applications [1,2]. General consensus is that with the use of ceramic components, thermal efficiency and functional lifetimes of engine systems can be improved substantially [3-5]. However, recent studies indicate that despite high expectations, most ceramics are not really very resistant to wear and their sliding friction coefficients against most counterface materials are rather high [2,3,6-8]. It is believed that without the development of advanced lubricants and lubrication concepts, ceramic materials cannot be used in advanced tribosystems [7,8].

To achieve low friction and low wear in ceramic materials, a number of approaches have emerged in recent years. These include liquid lubrication, vapor-phase-lubrication, and solid lubrication. Liquid lubrication of ceramic surfaces is much desired but suffers from the inability of existing lubricants to function efficiently over wide temperature ranges [9,10]. When used at elevated temperatures, conventional lubricants may break down or undergo severe oxidation. Under extreme cases, a layer of undesirable deposits may form on the rubbing surfaces of tribocomponents. Furthermore, conventional additives used in these lubricants do not appear to respond well to most ceramic surfaces. Consequently, boundary films, often necessary for extreme contact situations, may not be formed on the sliding surfaces of ceramic materials [9]. To overcome these shortcomings, new liquid lubricants and additives are being developed and presently tried on ceramic surfaces [10].

Vapor-phase lubrication of ceramic tribocomponents has lately attracted a great deal of attention from researchers involved in advanced heat-engine research. Initial results from a series of field tests appear very promising [11]. Thin carbon films resulting from catalytic reactions between carbonaceous gases and ceramic surfaces were also found to result in low friction at elevated temperatures [12].

As for solid lubricants, their fabrication as thin adherent coatings on ceramic surfaces has so far been very difficult with conventional deposition methods. However, by using such advanced methods as ion-beam-mixing [13,14] and ion-beam-assisted deposition (IBAD) [15-17], very adherent lubricious coatings have been deposited successfully on ceramic substrates. For example, using an IBAD system, we achieved the deposition of very adherent Ag coatings on Al_2O_3, ZrO_2, Si_3N_4, and sapphire substrates [15,18,19]. Moreover, we recently demonstrated that through the use of adherent Ag coatings, both the friction and wear properties of these ceramics can be substantially improved. For the deposition of these initial films, we used Ar ion beams emitted from a Kaufman-type ion source. Since these initial studies, we have found by experimentation that through the use of an ion flux consisting of a 1:1 mixture of argon and oxygen, even greater adhesion can be achieved between silver films and ceramic substrates [17,19]. As a matter of fact, using a pull-type adhesion test device, we were unable to separate these Ag coatings from the substrates. Figure 1, based on the data reported in ref. 17, compares the adhesive strength of a series of Ag coatings grown under various deposition conditions. As is clear, the use of an Ar+O ion flux make a significant difference in the adhesion of resultant Ag films. Therefore, in this paper, we will present the results of the recent tribological tests conducted on these more adherent Ag coatings. To explore their potential for tribological applications, gold (Au) coatings were also deposited on Al_2O_3 ceramics and tested under the same conditions as those for Ag coatings.

EXPERIMENTAL DETAILS

Material

The flat alumina specimens, with nominal dimensions of 38.1 x 50.8 x 4.2 mm, average surface roughness of about 0.3 μm center-line-average (CLA) and grain sizes ranging from 10 to 20 μm were obtained from a commercial company. According to the technical specifications of this company [20], Al_2O_3 specimens were fabricated from high-purity α-Al_2O_3 powders by means of sintering in air at ≈1800°C. 12.7-mm-diameter Al_2O_3 balls with a surface finish of better than 0.01 μm CLA were used as counterparts in the wear test machine. The volume porosity of the test materials was less than 1% and the bulk Vickers hardness was about 16 GPa. Prior to wear testing and/or film deposition, the specimens were ultrasonically cleaned sequentially in hexane + 10% toluene, acetone, deionized water + 2% laboratory detergent, and deionized water for 1 min. each, then oven-dried at 110°C for 20 min. This cleaning procedure was shown to be very effective in removing most of the adsorbed contaminants from the surface of Al_2O_3 ceramics [21].

Ion-beam-assisted deposition of Ag and Au

Ag and Au coatings were produced on Al_2O_3 substrates in an ion-beam-assisted-deposition system incorporating a dual electron-beam-heated evaporator. A schematic depiction of the IBAD system is shown in Figure 2. The coating materials, Ag and Au, were chosen mainly because of their excellent thermal and chemical stabilities over wide temperature ranges. They are not expected to diminish due to oxidative wear. Furthermore, they possess very high thermal conductivities, hence they can help reduce thermal and/or thermomechanical instabilities that occur at sliding interfaces of ceramics mainly because of frictional heating. Such instabilities are known to cause severe wear damage on ceramics, especially on those possessing poor thermal conductivities [19].

In the past, Ag and Au coatings have been used to lubricate vacuum tube and aerospace bearings. Compared to Au, Ag is relatively cheap and has served as an essential ingredient in the fabrication of the PS-200 (a plasma sprayed self-lubricating coating) [22] and the PM-200 (a powder metallurgical self-lubricating composite) [23] series of self-lubricating composite materials developed by NASA scientists.

Ion bombardment of Ag and Au coatings was done with a 1:1 mixture of Ar and O ions produced by a hot-cathode Kaufman-type ion gun. This type of deposition regime was found earlier to be very effective in affording strong adhesion between Ag coatings and Al_2O_3 substrates [17]. Ions emitted from the ion source were neutralized by a hot-wire filament, thereby reducing the charging of insulating Al_2O_3 substrates. Prior to the deposition of Ag and Au, the substrates were sputter-cleaned using the same ion flux. Typical base pressures in the turbopumped IBAD system were approximately 10^{-5} Pa prior to and about 10^{-2} Pa during deposition. The evaporation rate was measured and controlled during the IBAD of Ag and Au using a quartz-crystal rate monitor. The thickness of the coatings used in our study was approximately 2 μm. Further details of the IBAD system and the structural and adhesion characteristics of coatings can be found in Refs. 15,19, and 24.

Friction and Wear tests

Friction and wear tests were performed with pairs of Al_2O_3 balls and Al_2O_3 flats, and Al_2O_3 balls and IBAD-Ag- and Au-coated Al_2O_3 flats in an oscillating ball-on-flat-type wear test machine. A schematic of this test machine is shown in Figure 3. Typically, the ball specimens were drawn back and forth across the stationary flats at a frequency of 1 Hz and for a stroke length of 25 mm. This results in an average sliding velocity of 0.05 m/s. In this study, three types of wear tests were performed. The standard tests, e.g., 4000 sliding cycles (8000 passes), were run to elucidate specifically the effect of increasing ambient temperature on the friction and wear behavior of Al_2O_3 with and without an IBAD-

Ag and/or Au coating. The long-range, e.g. 12,000 sliding cycles (24,000 passes), and the lifetime wear tests, e.g., 55,000 cycles (110,000 passes), were performed mainly to assess the durability and long-term lubrication capability of Ag- or Au-coatings produced on ceramic substrates.

The 12.7-mm diameter Al_2O_3 balls were firmly secured on a holder to assume the oscillating-ball configuration of the oscillating ball-on-flat test rig (see Fig. 3). The dead weight on top of the oscillating ball was maintained at 5 N which created an initial mean Hertzian contact pressure of about 0.63 GPa on the uncoated flats. Because of the progressive wear of the Al_2O_3 balls, the contact area increased and the initial contact stress decreased sharply to a value that is a very small fraction of the initial mean Hertzian contact pressure (See nominal pressure values in figures showing the wear data).

The standard wear tests were conducted at room temperature, 200, and 400^0C. The elevated test temperatures were created with a series of electrical cartridge heaters inserted into the top and the bottom stainless-steel enclosures that housed the ball and flat specimens. Further details of this tester can be found in Ref. 16. All tests were performed under unlubricated sliding conditions in open air of 20 ± 4% relative humidity. Frictional force was monitored with the aid of a strain-gauge transducer. The measurement of wear volume for the balls was based on the microscopic determination of the wear-scar diameter, which was subsequently converted into wear volume by a mathematical expression suggested in ref. 25. The wear of flats was assessed with the aid of a surface profilometer. Two to three duplicate tests were run at each temperature and the mean values of friction coefficients and wear rates are reported in appropriate graphs.

RESULTS

Figure 4 shows the average wear rates of the uncoated balls slid against the uncoated and Ag- and Au-coated flats as a function of increasing temperature. Values in parentheses indicate the final nominal contact pressures developed between balls and flats. According to the data, without a coating, the average wear rates of both the balls and the flats increase by a factor of about 10 when temperature is raised to 200°C. Specifically, the average wear rate of uncoated flats is about 2.2 x 10^{-5} $mm^3.N^{-1}.m^{-1}$ at room temperature, but increases to 2.0 x 10^{-4} $mm^3.N^{-1}.m^{-1}$ at 200°C. For the uncoated balls, the average wear rate is about 0.5 x 10^{-5} $mm^3.N^{-1}.m^{-1}$ at room temperature but goes up to 4.5 x 10^{-5} $mm^3.N^{-1}.m^{-1}$ at 200°C. When ambient temperature is further increased to 400°C, the wear rate of flats remains relatively unchanged, but the wear rate of balls increase slightly.

Electron microscopy of the wear scars and tracks formed on uncoated balls and flats revealed extensive cracks often associated with internal voids, flaws, and weak grain-boundaries (see Figure 5). At 200 and 400°C, the dominant cause of wear appeared to be still microfracture (see Fig. 6a). Some microfeatures suggesting plastic flow were also noticed on the worn surfaces of uncoated balls and flats as seen in Fig. 6b. However, at higher magnifications, it became clear that these features were actually due to compacted and/or smeared wear-debris particles with sizes ranging from 10 to 100 nm.

The wear rate of the Al_2O_3 balls sliding against the IBAD-Ag or Au-coated Al_2O_3 flats was about 1.2 x 10^{-7} $mm^3.N^{-1}.m^{-1}$ at room temperature. However, at 200°C, the wear rate of balls was reduced slightly to 0.8 x 10^{-7} $mm^3.N^{-1}.m^{-1}$, whereas, at 400°C, the wear rate of the balls increased to about 2.5 x 10^{-7} $mm^3.N^{-1}.m^{-1}$. The wear rates of the IBAD-Ag- and Au-coated Al_2O_3 flats could not be assessed by a surface profilometer. As evident from the SEM micrographs in Figure 7, these coatings remained largely intact on the rubbing surfaces. However, because of their soft nature, some plastic deformation occurred during dynamic sliding. As is evident from Figure 7b, only the tips of underlying asperities were exposed and slightly worn away (appearing as dark regions in micrographs).

The steady-state friction coefficients of various test pairs are summarized in Table 1. As is clear, the friction coefficients of pairs without coatings are quite high (e.g., 0.8 at room temperature) and increase substantially with increasing temperature (e.g., 1.1 at 400°C), whereas, the friction coefficients of pairs with an IBAD-Ag or Au coating are generally low, typically ranging from 0.35 to 0.5.

Figure 8 presents the wear performance of balls subjected to long-range wear tests at room temperature. Again, the values in parentheses indicate the final nominal contact pressures between balls and flats. The wear of IBAD-Ag- and Au-coated flats was practically unmeasurable with a profilometer, and the wear of uncoated balls sliding against the Ag- or Au-coated flats was reduced by more than two orders of magnitude. Microscopic inspection of the wear tracks revealed that even after many sliding cycles, Ag and Au remained still intact on the wear tracks of these flats.

As for the lifetime tests (e.g. 55,000 cycles at room temperature), the average wear rates of balls slid against the Ag- and Au-coated flats were 4.3×10^{-8} $mm^3.N^{-1}.m^{-1}$ and 4.9×10^{-8} $mm^3.N^{-1}.m^{-1}$, respectively. In contrast, the average wear rates of balls slid against the uncoated flats for the same amount of sliding was in the order of 1.6×10^{-6} $mm^3.N^{-1}.m^{-1}$. These lower average wear rates for the lifetime tests are consistent with the expected steadily decreasing instantaneous wear rates. As is evident from the micrographs in Figure 9, Ag and Au films are still present within the wear tracks. In all cases, it was found that some Ag and Au had transferred to and around the wear scars formed on ball counterfaces (see Figure 10).

DISCUSSION

The results presented above demonstrate that both the wear rates and the friction coefficients of self-mated Al_2O_3 ceramics are quite high at room temperature and tend to increase with increasing temperature (see Figure 4 and Table 1). These observations are consistent with those reported by Woydt and Habig [26], Erdemir et al. [16], and Yust and Carignan [27] who also measured significantly high wear rates and friction coefficients for self-mated pairs of Al_2O_3 ceramics tested at elevated temperatures. In contrast, the wear rates of IBAD-Ag- or Au-coated Al_2O_3 flats were unmeasurably low and the wear rates of counterface Al_2O_3 balls sliding against these flats were reduced by factors of 45 at room temperature to more than 500 at 200°C. Furthermore, friction was also reduced to about one-half to one-third that of uncoated pairs (see Table 1).

For the high wear rates of uncoated flats and balls, we propose the following explanation. It is known that both the magnitude and the location of subsurface stress fields in sliding contacts are strongly related to the extent of friction coefficient [28]. According to Hamilton and Goodman [28], when the friction coefficient exceeds 0.3, the tangential traction between sliding pairs is substantially increased and the location of maximum shear forces is shifted from subsurface to the sliding surface. As a rule, higher the friction coefficient, higher the magnitude of shear force and its tensile component developing behind the moving surface asperities. As shown in Table 1, the steady-state friction coefficient of Al_2O_3 on itself is quite high and tends to increase with temperature. Therefore, under the test conditions of this study, the maximum shear stresses were always at the plane of sliding contacts and that large tensile forces developed behind the moving surface asperities.

Because of its inherent brittleness, Al_2O_3 cannot sustain large strains when pulled in tension. It prematurely undergoes brittle fracture. We believe that under the influence of surface tensile stresses, microcracks initiate primarily from such volume defects as micropores, flaws, secondary phases, and discontinuous grainboundaries lying at or in the vicinity of contact interfaces. As is known, internal volume defects and grain boundaries can act as stress concentrators and lower the fracture strength of ceramics, especially under concentrated loading conditions [29]. Due to the cyclic nature of oscillating sliding, in the long run, surface fatigue may also play some roles in the initiation of microcracks and promotes wear [16,27]. SEM micrographs in Figure 5 and 6b support our interpretation by revealing extensive microcracks (both inter- and intra-granular types) on the rubbing surfaces of Al_2O_3 samples.

As shown in Figure 4, the wear rates of uncoated balls and flats increase with increasing temperature. We believe that at elevated temperatures, the magnitude of the tensile forces which develop behind the moving surface asperities was much greater than at room temperature, because, as shown in Table 1, the friction coefficients are somewhat higher at elevated temperatures than at room temperature. Larger tensile forces developing behind the moving surface asperities are thought to produce greater microfracture, hence higher wear.

As was reported in Refs. 14, 26, and 28 the hardness of Al_2O_3 decreases substantially with increasing temperature. We do not know the exact role of decreasing hardness in increasing wear rates of Al_2O_3 at present. However, it is possible that the initiation of microcracks from internal volume defects (e.g., porosities, flaws) may have become much easier at elevated temperatures. Furthermore, intergranular debonding due to the weakining of grainboundaries and/or grainboundary precipitates may have been augmented [29].

As shown in Figure 4 and Table 1, IBAD-Ag and -Au films can dramatically reduce the wear rates of Al_2O_3 ceramics. The friction coefficients are also substantially reduced. The amount of reduction is more pronounced at elevated temperatures than at room temperature. From these observations, we believe that to achieve low friction and low wear in Al_2O_3 ceramics, sliding surfaces must be supplied with a thin film that can shear readily and continuously. Furthermore, the film must be thick enough to prevent opposing asperities from coming into frequent contacts. Our results demonstrate that the adherent Ag and Au films have the ability to effectively accomodate surface tensile stresses by shearing easily. Easy shear at contact interfaces is analogous to low friction, whereas fewer asperity/asperity interaction means lower wear. When friction is low, the magnitude of detrimental tensile stresses developing behind the moving surface asperities, hence the probability of microcrack initiation on sliding surfaces are reduced [28]. Because of a larger load-bearing area afforded by the easily-deforming Ag and Au films (see Figs. 7 and 9), the magnitude of normal forces is also reduced. This further helps reduce the probability of crack initiation especially from those internal volume defects acting as stress concentrators.

As is evident from Figures 7 and 9, despite high contact stresses and several thousands of stress cycles, both Ag and Au coatings remained intact on the sliding surfaces of Al_2O_3 substrates. In most cases, only the tips of original surface asperities are exposed and somewhat worn away, but the whole surface was well-protected.

For the long endurance lives of Ag and Au coatings, (see Fig. 8) we propose two mechanisms: (1) transfer and back-transfer of coating material from one sliding surface to another (see Fig. 10) and (2) strong film-to-substrate adhesion imparted by the IBAD process. According to Lancaster [30], the transfer of coating material from one face to the other tends to keep the coating material within the sliding interface, thereby increasing the functional lifetime of these coatings. However, such a transfer and back-transfer may have an adverse effect on friction. Because, according to the adhesion theory of friction, the formation and subsequent rupture of microjunctions between original and transferred film material can give rise to higher friction. As apparent from Table 1, the range of friction coefficients, e.g., 0.35 to 0.5, is indeed somewhat high and may be impractical for some practical applications.

In recent studies, we found that (1) without adequate adhesion, Ag coatings were removed rapidly from the sliding surfaces of ceramic substrates and (2) through the use of concurrent ion bombardment during film formation, the adhesion of Ag coatings was markedly improved. Additionally, we have found that presputtering of Al_2O_3 substrates with either oxygen ions or a combination of argon and oxygen ions results in even much stronger adhesion between Ag films and ceramic substrates [17]. Using a pull-type adhesion test machine, we could not separate these Ag films from substrates. Detachment has always occurred between the adhesive glue and the surface of coatings. Adhesive glue had a tensile strength of about 70 MPa, which means that adhesive strength of surface coatings was greater than 70 MPa.

Although we do not yet know the precise role of O ions in improved coating adhesion, we believe that, with the use of O ion bombardment, the surface chemical reactivity of Al_2O_3 is increased. As a result, greater chemical bonding is achieved across the interface between silver film and Al_2O_3 substrate. Several other mechanisms have been proposed for the excellent adhesion of coatings produced under ion bombardment [31]. Some of these are: establishment of a graded shallow interface, mechanical interlocking of coating and substrate materials at the atomic level, and greater nucleation density of the IBAD-produced films [31].

CONCLUSIONS

* Through the use of Ar + O ion beams, very adherent Ag and Au coatings can be produced on ceramic substrates. These coatings are capable of virtually eliminating the wear of Al_2O_3 substrates. The wear rates of counterface balls are also reduced by factors of 45 at room temperature to more than 500 at 200^{oC}.

* Dry sliding of uncoated Al_2O_3 balls against uncoated flats results in severe wear damage and high frictional losses. Increasing temperature has a detrimental effect on the friction and wear behavior of Al_2O_3. Microfracture is the dominant cause of wear.

* IBAD- Ag and Au coatings remain largely intact on the wear tracks of Al_2O_3 flats even after 100,000 sliding passes. Some Ag and Au may transfer to the rubbing surfaces of counterface balls.

* When these ceramics are considered for use in high-temperature tribosystems, their high friction coefficients and high wear rates should be recognized as major shortcomings. As demonstrated in our study, the use of lubricious coatings may be helpful. In particular, a liquid-solid lubrication approach can be of significant help in controlling the wear of ceramic materials over wide ranges of operating temperature [32].

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Table 1. Friction coefficients of various test pairs.

Test Pairs	Steady-state Friction Coefficient		
Ball/Flat	R.T.	200°C	400°C
Al_2O_3/Al_2O_3	0.8	1.0	1.1
Al_2O_3/IBAD-Ag-Coated Al_2O_3 Flat	0.5	0.35	0.40
Al_2O_3/IBAD-Au-Coated Al_2O_3 Flat	0.35	0.32	0.45

FIGURE CAPTIONS:

Fig. 1. Adhesion strength of Ag coatings produced under various deposition conditions.

Fig. 2. A schematic depiction of the ion-beam-assisted-deposition system.

Fig. 3. A schematic illustration of the oscillating-slider wear test machine.

Fig. 4. Variation of wear rates of uncoated disks and of balls sliding against silver- and gold-coated flats at different test temperatures (values in parentheses indicate final nominal contact pressures in MPa).

Fig. 5. SEM micrograph of a wear track formed on an Al_2O_3 flat during sliding against an Al_2O_3 ball at room temperature.

Fig. 6. SEM micrographs of wear tracks formed on Al_2O_3 flats after testing at (a) 200°C and (b) 400°C. Features suggesting plastic deformation in (b) are due to compacted and/or smeared wear-debris particles.

Fig. 7. Low-magnification back-scattered SEM and b) high-magnification secondary SEM images of a wear track formed on an IBAD-Au-coated Al_2O_3 flat during a standard wear test at 200°C.

Fig. 8. Wear performance of Al_2O_3 balls sliding against uncoated and IBAD-Ag- and Au-coated flats in long-range wear tests (values in parentheses indicate final nominal contact pressures in MPa).

Fig. 9. SEM micrograph of the wear track of (a) an IBAD-Ag-coated and (b) an IBAD-Au-coated Al_2O_3 flat after sliding 55,000 cycles.

Fig. 10. (a) Secondary and (b) back-scattered SEM micrographs of a wear scar formed on an Al_2O_3 ball during lifetime test with a silver-coated Al_2O_3 flat.

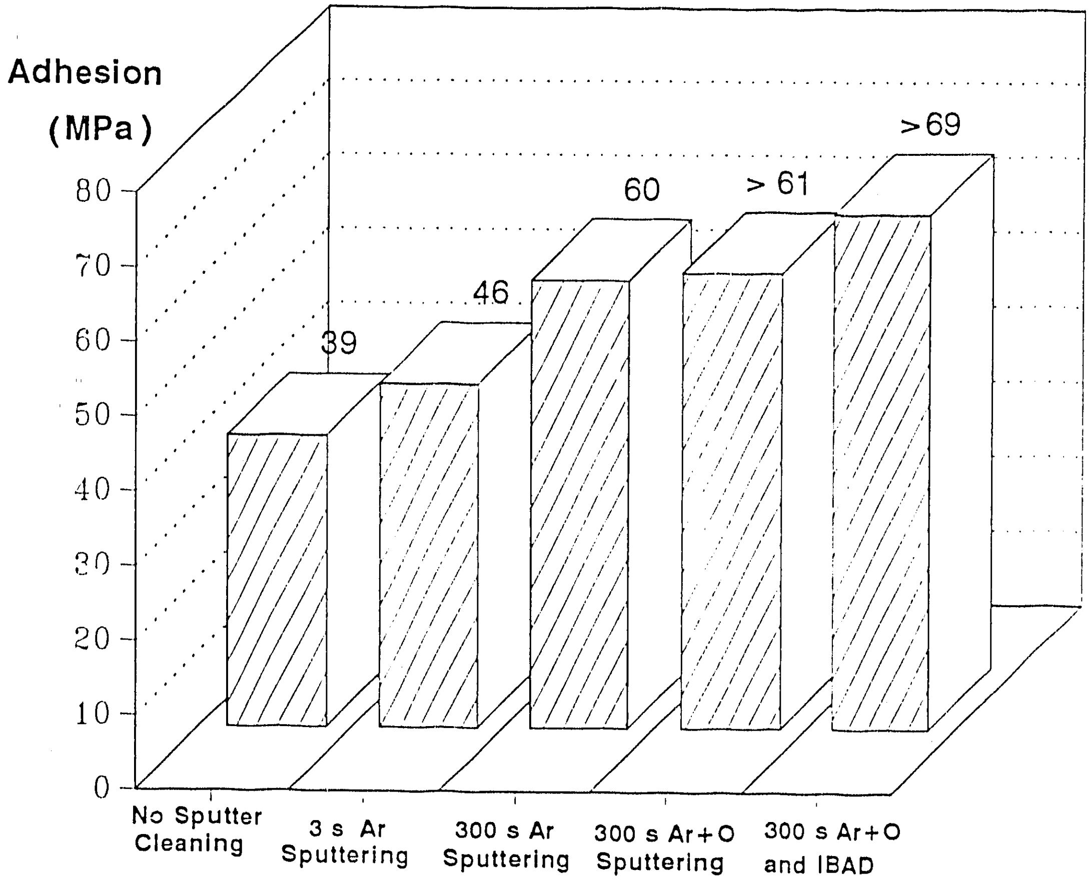


Fig. 1. Adhesion strength of Ag coatings produced under various deposition conditions.

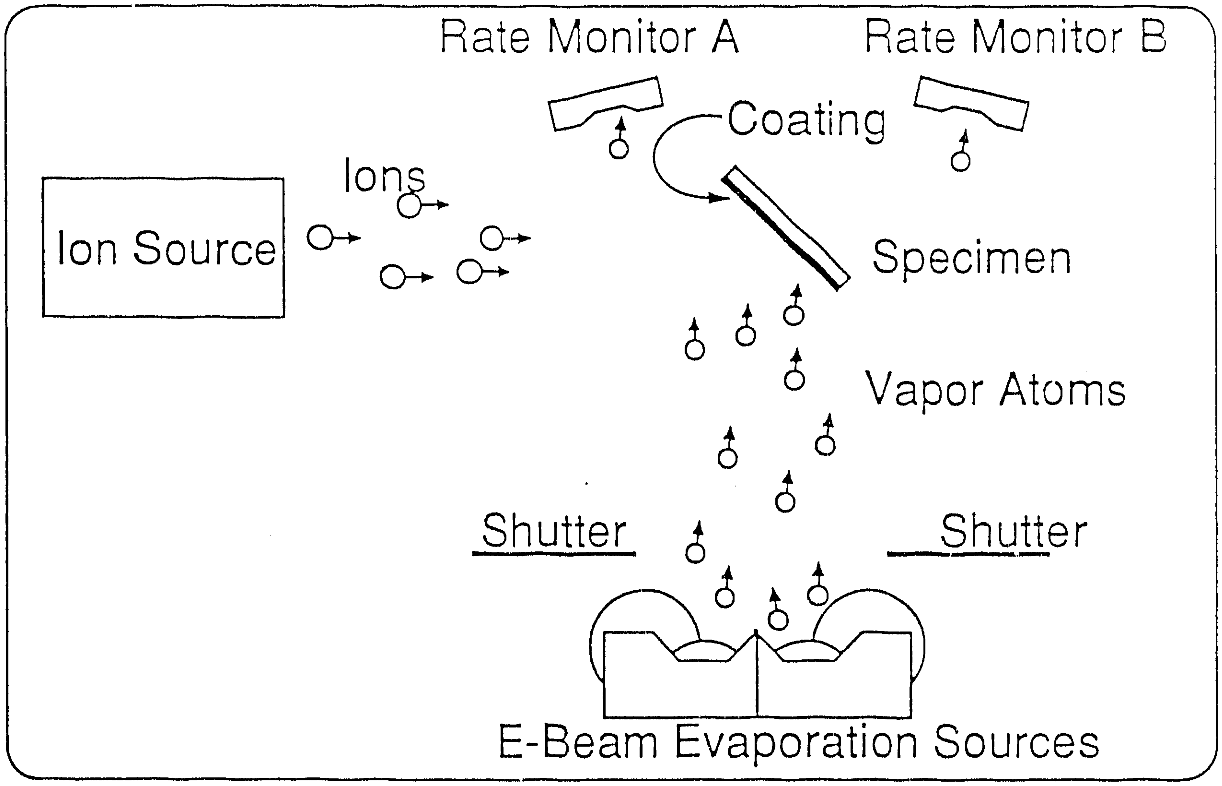


Fig. 2. A schematic depiction of the ion-beam-assisted-deposition system.

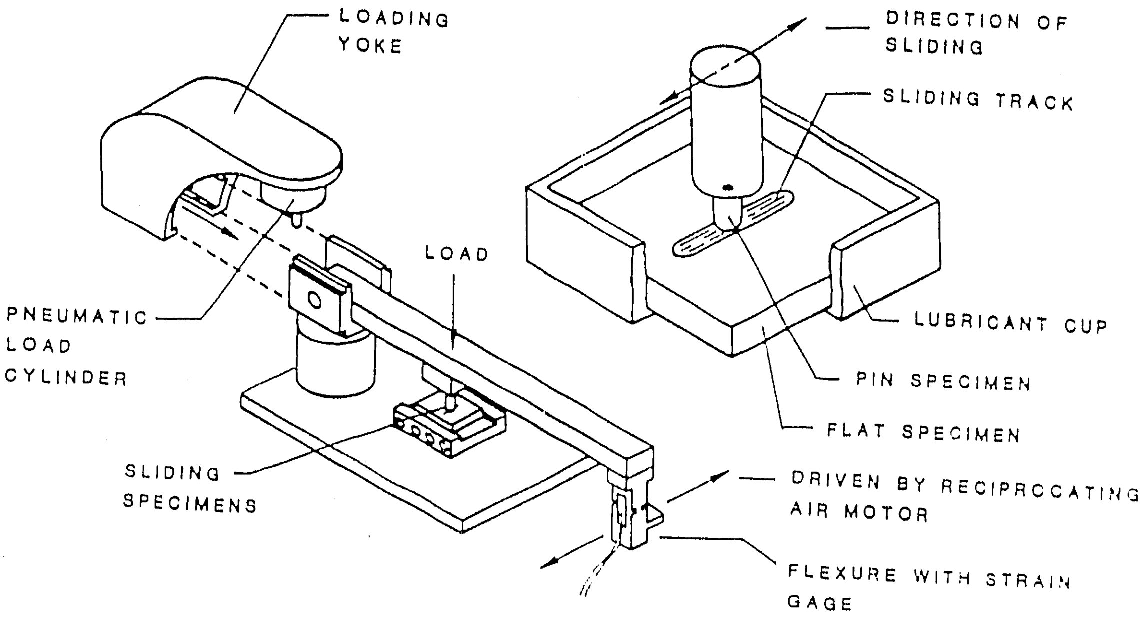


Fig. 3. A schematic illustration of the oscillating-slider wear test machine.

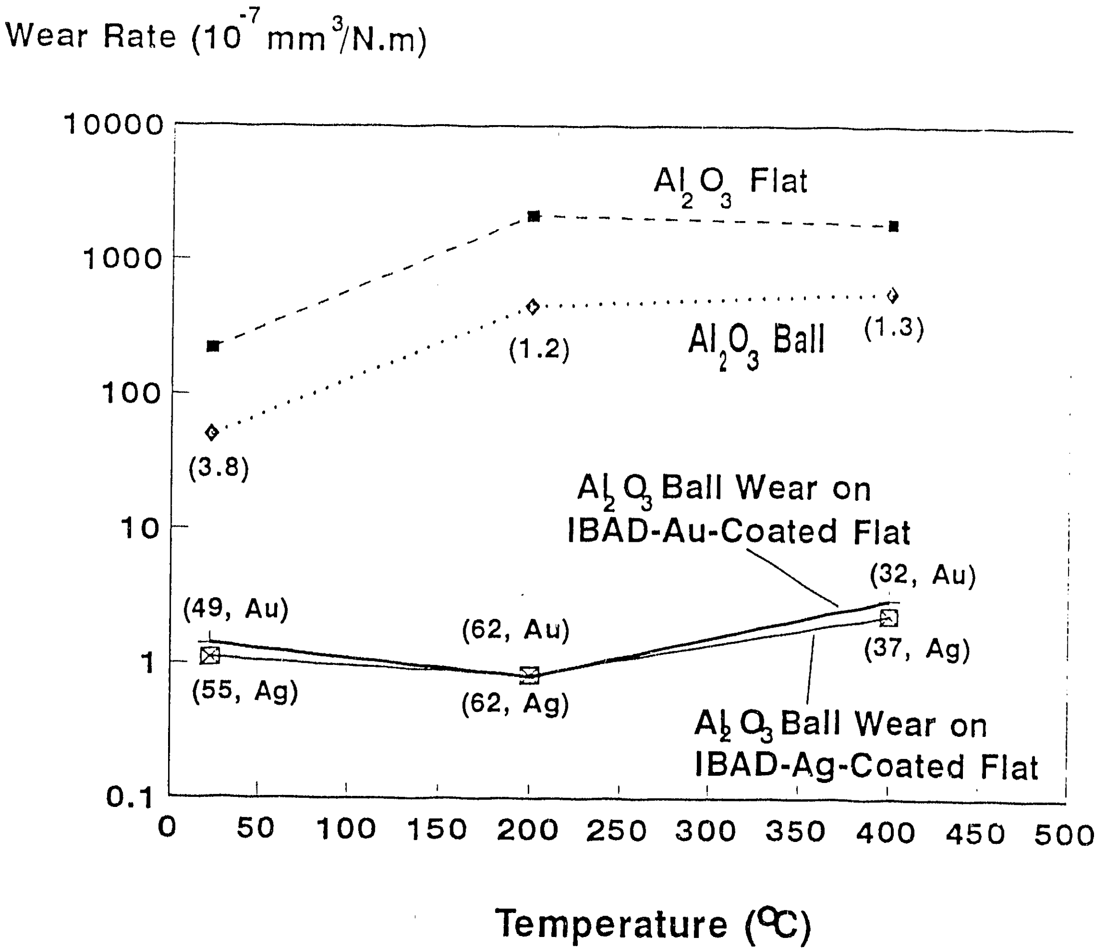
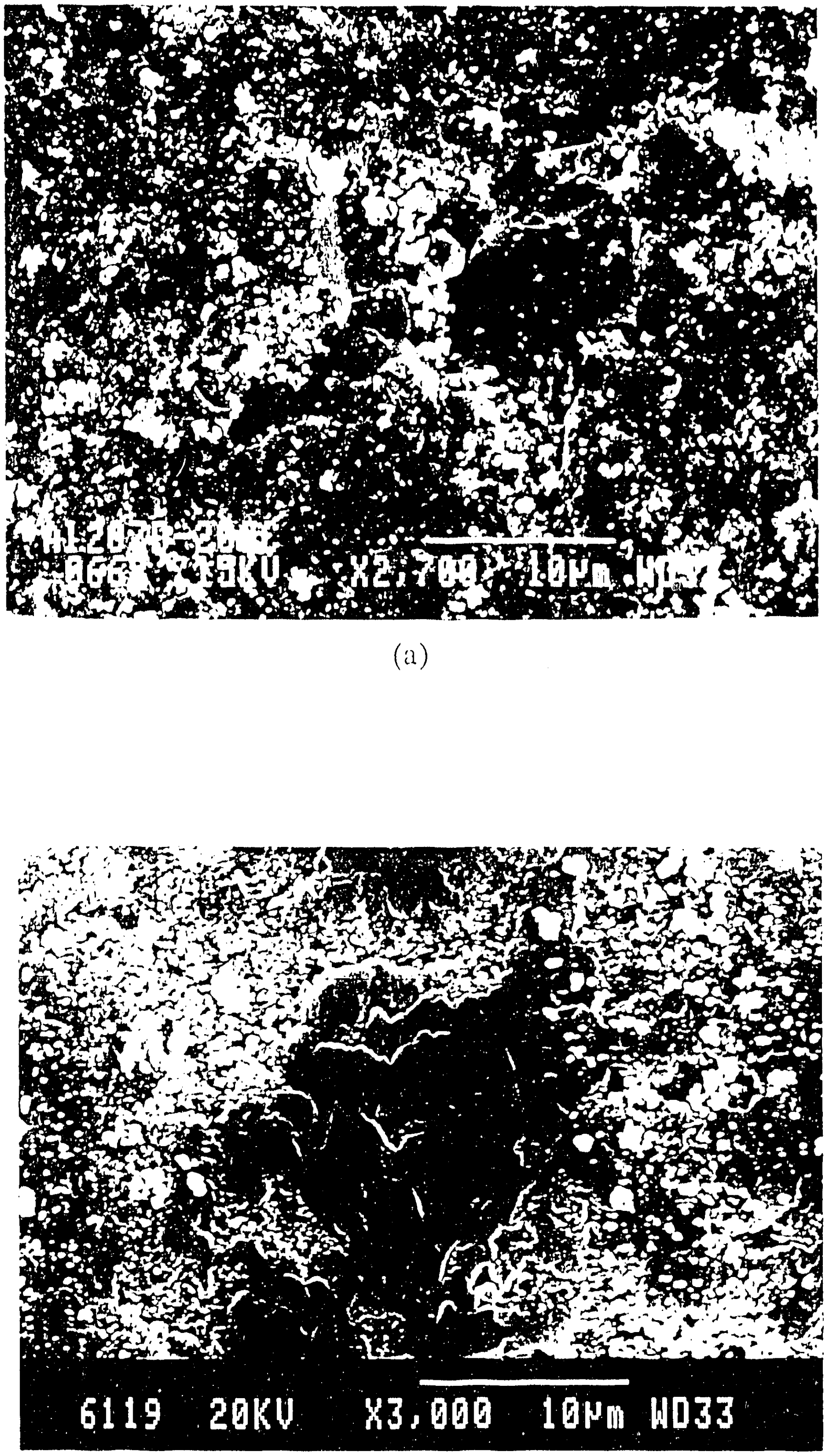


Fig. 4. Variation of wear rates of uncoated disks and of balls sliding against silver- and gold-coated flats at different test temperatures (values in parentheses indicate final nominal contact pressures in MPa).

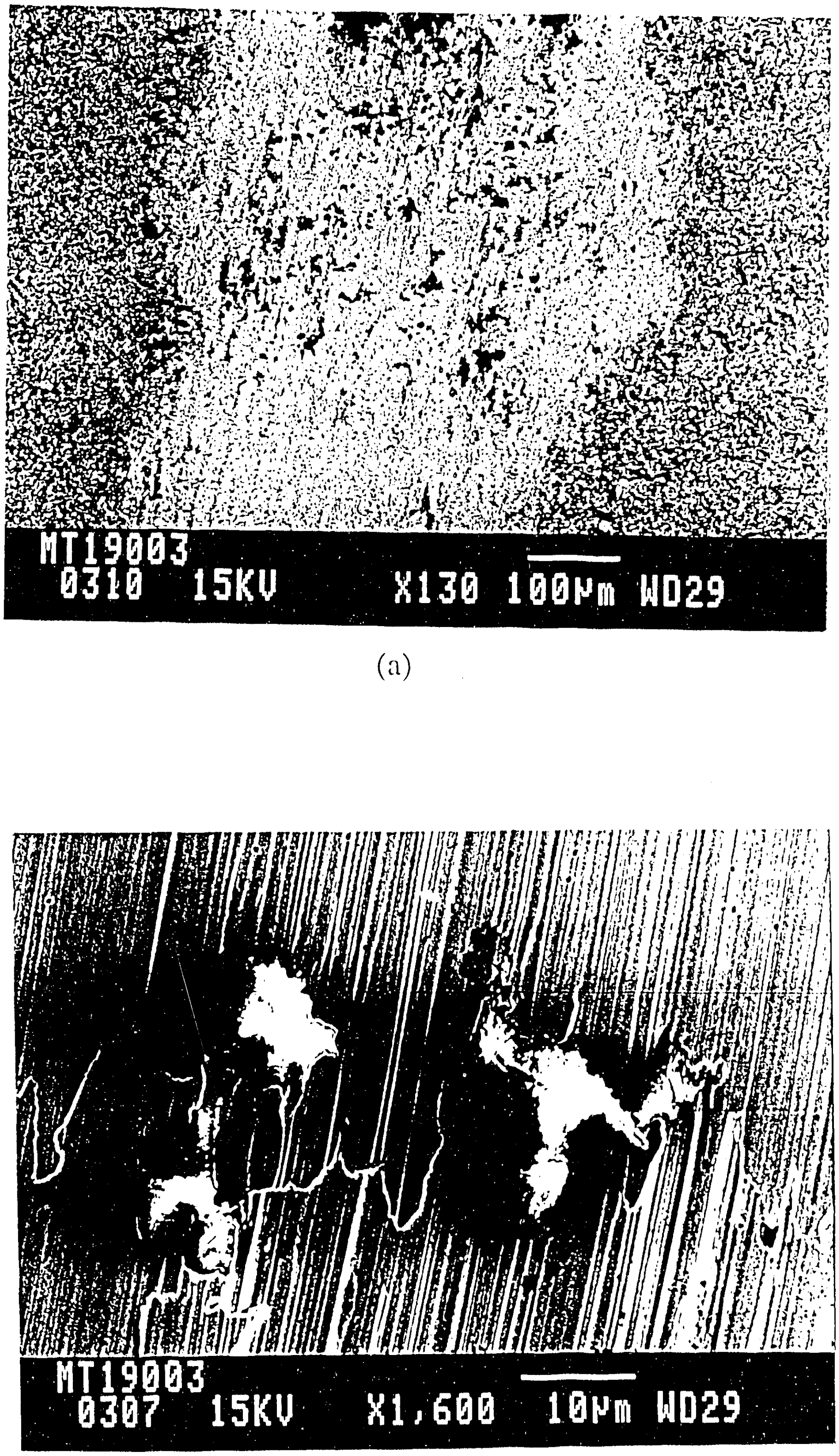
Fig. 5. SEM micrograph of a wear track formed on an Al_2O_3 flat during sliding against an Al_2O_3 ball at room temperature.



(a)

(b)

Fig. 6. SEM micrographs of wear tracks formed on Al_2O_3 flats after testing at (a) 200°C and (b) 400°C. Features suggesting plastic deformation in (b) are due to compacted and/or smeared wear-debris particles.



(a)

(b)

Fig. 7. a) Low-magnification back-scattered SEM and b) high-magnification secondary SEM images of a wear track that formed on an IBAD-Au-coated Al_2O_3 flat during a standard wear test at 200°C.

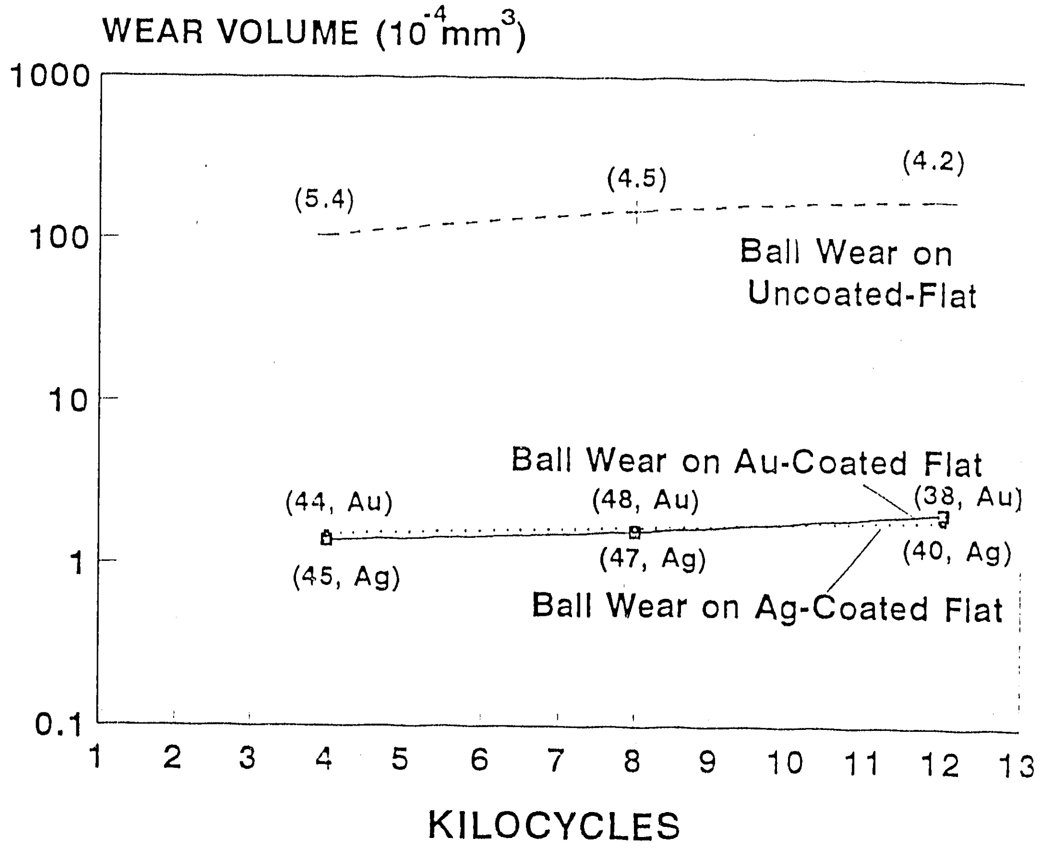
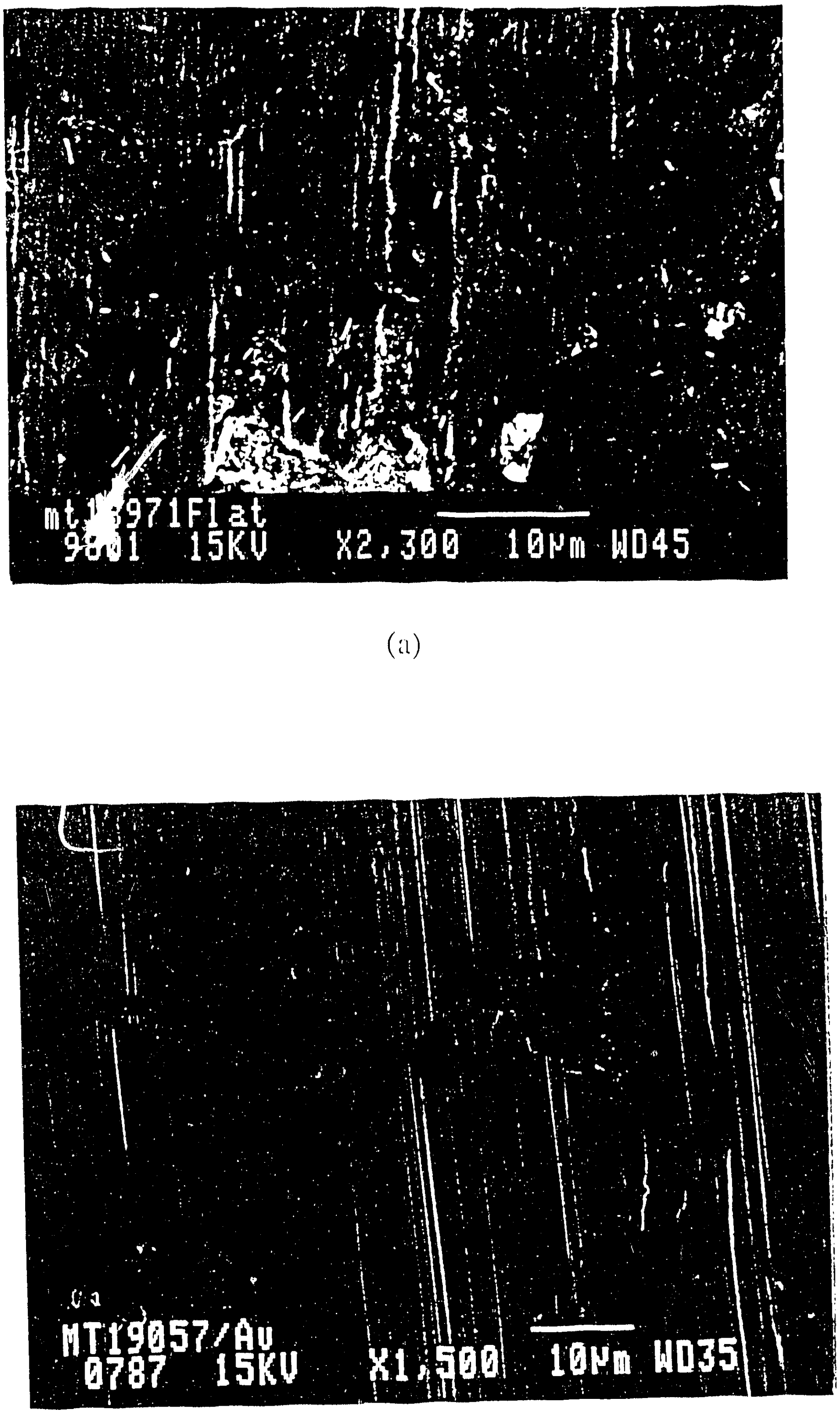


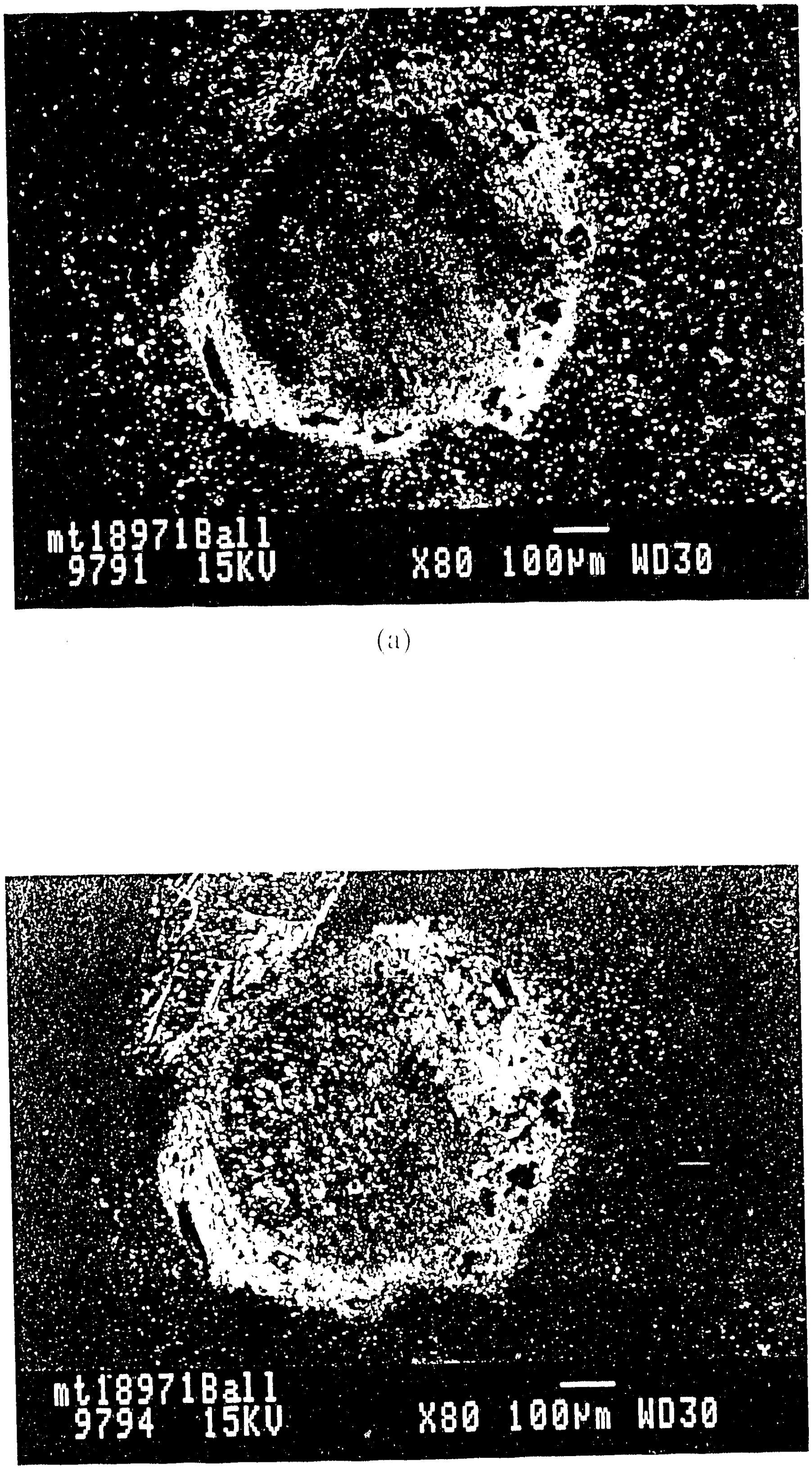
Fig. 8. Wear performance of Al_2O_3 balls sliding against uncoated and IBAD-Ag- and Au-coated flats in long-range wear tests (values in parentheses indicate final nominal contact pressures in MPa).



(a)

(b)

Fig. 9. SEM micrograph of the wear track of (a) an IBAD-Ag-coated and (b) an IBAD-Au-coated Al_2O_3 flat after sliding 55,000 cycles.



(a)

(b)

Fig. 10. (a) Secondary and (b) back-scattered SEM micrographs of a wear scar formed on an Al_2O_3 ball during lifetime test with silver-coated Al_2O_3 flat.

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