

Deposition of various metal, ceramic, and cermet coatings by an industrial-scale large area filtered arc deposition process

V. Gorokhovskiy,^{a)} C. Bowman, D. VanVorous, and J. Wallace
Arcomac Surface Engineering, LLC, Bozeman, Montana 59715

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Nearly defect-free nitride, carbide, and oxiceramic coatings have been deposited by a unidirectional dual large area filtered arc deposition (LAFAD) process. One LAFAD dual arc vapor plasma source was used in both gas ionization and coating deposition modes with and without vertical magnetic rastering of the plasma flow. Substrates made of different metal alloys, as well as carbide and ceramics, were installed at different vertical positions on the 0.5 m diameter turntable of the industrial-scale batch coating system which was rotated at 12 rpm to assess deposition rates and coating thickness uniformity. Targets of the same or different compositions were installed on the primary cathodic arc sources of the LAFAD plasma source to deposit a variety of coating compositions by mixing the metal vapor and reactive gaseous components in a magnetically confined, strongly ionized plasma flow with large kinetic energy. The maximum deposition rate typically ranged from 1.5 $\mu\text{m/h}$ for TiCr/TiCrN to 2.5 $\mu\text{m/h}$ for Ti/TiN multilayer and AlN single layer coatings, and up to 6 $\mu\text{m/h}$ for AlCr-based oxiceramic coatings for primary cathode current ranging from 120 to 140 A. When the arc current was increased to 200 A, the deposition rates of TiN-based coatings were as high as 5 $\mu\text{m/h}$. The vertical coating thickness uniformity was $\pm 15\%$ inside of a 150 mm area without vertical rastering. Vertical rastering increased the uniform coating deposition area up to 250 mm. The coating thickness distribution was well correlated with the output ion current distribution as measured by a multisection ion collector probe. Coatings were characterized for thickness, surface profile, adhesion, hardness, and elemental composition. Estimates of electrical resistivity indicated good dielectric properties for most of the TiCrAlY-based oxiceramic, oxinitride, and nitride coatings. The multielement LAFAD plasma flow consisting of fully ionized metal vapor with a reactive gas ionization rate in excess of 50% was found especially suitable for deposition of nanocomposite, nanostructured coatings. Potential industrial applications of this highly productive coating deposition process are discussed. © 2009 American Vacuum Society. [DOI: 10.1116/1.3114462]

I. INTRODUCTION

During the past 30 years, from the time when it was first introduced by the Kharkov Physical-Technical Institute research group, filtered cathodic arc deposition (FCAD) technology has slowly but surely made inroads into the world of physical vapor deposition (PVD).¹⁻³ The main advantage of the FCAD process is that it can generate a fully ionized metal vapor plasma consisting of atomic ions with a substantial concentration of multicharged ions but without contamination from multiatomic clusters and macroparticles. The filtered arc vapor plasma beam consists of a dense plasma with a relatively high electron temperature of the order of 3–5 eV; this contributes to an increased ionization rate of the gaseous atmosphere via electron collisions with gas atoms and molecules.

The kinetic energy of the FCAD metal ion flows exceeds the thermal energy of the FCAD plasma by an order of magnitude, ranging from 40 to 200 eV.^{1,3-6} In sharp contrast to sputtering processes, the cathodic arc evaporation process (both direct arc and filtered arc) does not require a minimum background pressure of inert gas and can generate metal va-

por plasma in any reactive gas atmosphere or even in deep vacuum.¹⁻⁵ These properties of the FCAD process make it advantageous compared to conventional- and even ionized-magnetron sputtering processes in many applications requiring high ionization rate of the depositing metal-gaseous vapor to increase coating density and other functional properties and to eliminate or substantially reduce the defects in the coatings.

The main obstacle to using conventional FCAD technology is the low productivity of this process, which restricts its usage to semiconductors, optical coatings, and some ultrathin hard coatings used in biomedical and tribological applications. On the other hand, large area filtered arc deposition (LAFAD) technology overcomes these limitations by providing a highly productive, robust, industry-friendly process that combines the high productivity rate of conventional direct cathodic arc deposition (DCAD) and magnetron technologies, with the capability of generating a nearly 100% ionized metal-gaseous vapor plasma with large kinetic energy and with no macroparticles, droplets, multiatom clusters, and other contaminants.^{4,7,8} The unidirectional dual arc LAFAD vapor plasma sources can be used as an alternative to conventional DCAD and magnetron based processes when the high productivity and uniformity needed for most industrial

^{a)}Electronic mail: vigase@aol.com

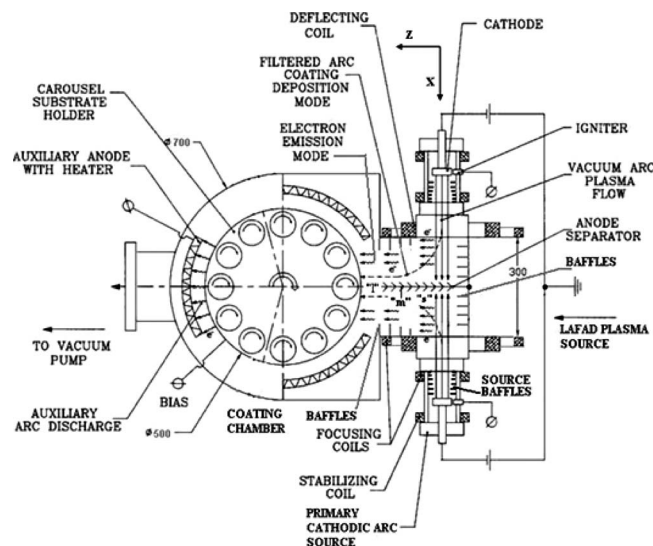


FIG. 1. Schematic illustration of the LAFAD vacuum system layout in this work (plan view). The unidirectional dual arc source assembly is to the right; the rotating substrate holder, equipped for dual rotation, is shown at the center of the drawing.

applications must be accompanied by the high ionization and high kinetic energy of atomically clean vapor plasma. Since the LAFAD plasma source operating pressure regimes are overlapping with most of the conventional vacuum vapor deposition technologies (magnetron sputtering, electron beam physical vapor deposition (EBPVD), thermal evaporation, plasma-assisted chemical vapor deposition (PACVD), etc.), it can be used in hybrid processes combining its high deposition and high ionization rates in conjunction with conventional PVD and low pressure PACVD processes, as was demonstrated in Refs. 8–10.

This work is dedicated to the characterization of the LAFAD process integrated into an industrial batch coater for the deposition of a wide variety of metal, ceramic, and cermet coatings of different compositions and architectures.

II. EXPERIMENTAL PROCEDURE

Figure 1 shows the LAFAD-500C batch coating system layout utilizing a vacuum chamber 0.7 m in diameter by 1 m tall, equipped with a unidirectional dual arc LAFAD plasma source. The rotatable substrate turntable, 0.5 m in diameter, is installed at the center of the coating chamber and allows for a single or double rotation of the substrates to be coated. The coating chamber is equipped with an array of radiant heaters and with diagnostic equipment including optical pyrometers and thermocouples to measure the substrate temperature. A set of sectional ion current collector probes can also be installed to measure ion current distribution in the vapor plasma stream. The LAFAD plasma source consists of the plasma guide chamber with baffles installed along its walls and the exit tunnel window which is 300 mm wide by 400 mm tall. Two pairs of deflection coils are located along the opposite walls of the plasma duct chamber. The primary DCAD sources, consisting of the cathode target, surrounded

by tubular anodes with steering and focusing coils, are attached to opposite walls of the plasma guide chamber adjacent to the exit tunnel portion. In addition, two vertical rastering coils are positioned on the top and bottom flanges of the plasma guide chamber for rastering the filtered arc flow.^{7–9} This provides high uniformity of the coating thickness distribution over large deposition areas.

When the deflection/focusing coils of the filter chamber are turned on, the vapor plasma generated by the primary DCAD sources flows into the plasma guide chamber from opposite directions and turns around the corner of the plasma guide exit tunnel toward the coating chamber. When the deflection/focusing coils of the filter chamber are turned off, an auxiliary arc discharge can be established between the primary arc cathodes of the LAFAD source and the auxiliary arc anode located in a coating chamber behind the turntable (see Fig. 1). This discharge provides ionization and activation of the gaseous atmosphere in the main chamber, producing a highly ionized gaseous plasma during such technological stages as ion cleaning/etching, gaseous ion implantation, and ionitriding/oxynitriding/carburizing.^{4,7–9,11}

An anode separator or dividing baffle can optionally be installed into the plasma duct to separate the two vapor plasma flows generated by the two primary cathodic arc sources.^{7,8} The dividing baffle can be either connected to the positive end of the arc power supply, grounded, or set up at floating potential which would be also positive with respect to the arc cathodes due to the higher mobility of the positive ions across the magnetized plasma confined in a longitudinal magnetic field. The experimental trials reported in this work were performed with and without the dividing baffles installed in the plasma guide chamber. Three dividing baffles with different lengths were used in this work: a short baffle, a medium length baffle, and a long baffle. The short baffle was installed between the back wall of the plasma guide chamber and a point 50 mm before the entrance of the tunnel section. The short dividing baffle end is marked “s” on the plan view of Fig. 1. The medium length baffle ends 50 mm into the tunnel section of the plasma guide chamber; its end is marked “m” in Fig. 1. The long baffle ends flush with the exit window of the LAFAD source; its end position is marked “l” in Fig. 1. A separation of the opposite vapor plasma flows generated by the two primary DCAD sources of the LAFAD plasma source allows the production of nanolaminated coatings by exposing the rotating substrates in turn to the plasma flows generated by opposite primary DCAD sources equipped with different targets (e.g., Ti and Cr, Ti and Al). When the dividing baffle is removed, the two opposite plasma flows generated by the primary DCAD sources with cathode targets of the same or different composition are mixed in the exit tunnel area, forming a uniform unidirectional plasma stream for the deposition of a wide variety of single component or multielemental nanocomposite coatings.

The sectional ion collector probes consist of either vertical or horizontal sections installed at the substrate turntable at a distance of 100 mm from the filter exit flange to measure the vertical and lateral ion current distributions in the FCAD

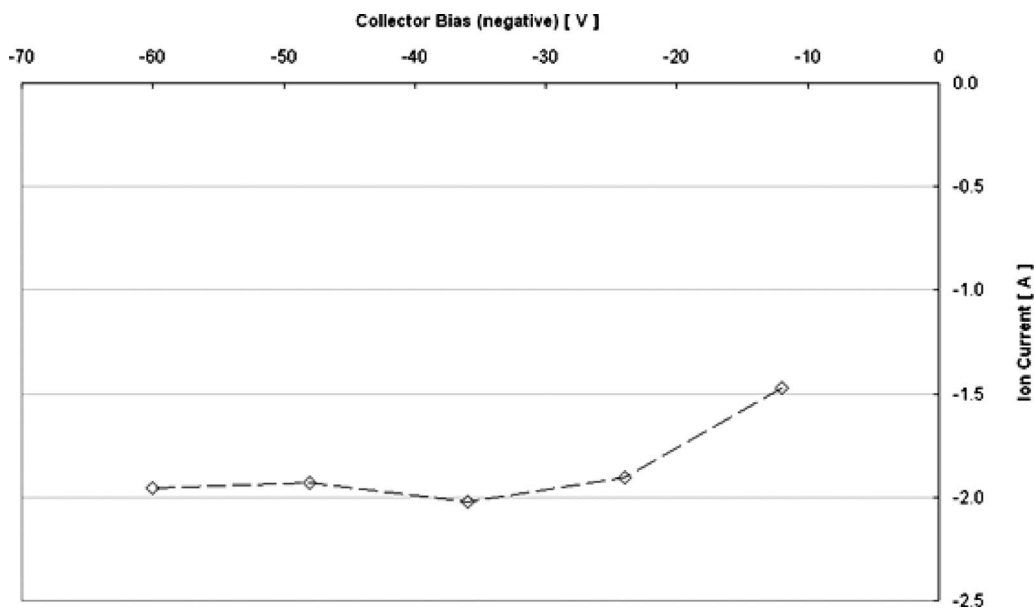


FIG. 2. Ion current as a function of ion collector bias. Saturation occurs for bias values beyond about -30 V.

vapor plasma flow similar to the methodology used in Ref. 12. In some cases, the ion collectors had only two sections covering the left and right sides of the exit window of the filter tunnel at a distance of about 20 mm from the tunnel exit to measure the integral ion currents propagating on the left and right sides of the tunnel as well as the total integral ion current yield transported through the filter. The ion probe power supply consisted of a set of 12 V lead-acid automotive batteries connected in series with a total voltage between 24 and 72 V dc in 12 V steps. The ion collector probes were connected to the negative end of the battery stack, while its positive end was connected to the cathode of one of the primary DCAD sources. Both titanium and carbon plasmas, generated by the primary DCAD sources with titanium and carbon targets, respectively, were used for ion current distribution measurements. Titanium filtered arc plasma includes a substantial concentration of multicharged ions ranging from Ti^+ to Ti^{4+} , while carbon filtered arc plasma consists of predominantly singly charged C^+ ions, which simplifies interpretation of ion current measurements.^{1,5,6} The calibration curve shown in Fig. 2 demonstrates a typical saturation Ti filtered arc ion current when the probe potential is reduced with respect to the reference cathode target potential; this shows that the ion current is fully saturated when the voltage drop between the cathode target and the collector probe is greater than or equal to -30 V.

The typical LAFAD plasma processing parameters used in the most of deposition trials reported in this work include the following steps: after preheating to 300 – 350 °C using a radiant heater array, the substrates were subjected to 15 or 20 min of ion cleaning at an Ar pressure of 0.8 mTorr in an argon auxiliary arc plasma generated by the LAFAD source with the deflecting magnetic field turned off. After this step, a short (2–3 min) high voltage metal ion-etching step was performed by subjecting the rotating substrates to 100% ion-

ized metal vapor plasma generated by the LAFAD source, while applying -800 to -1000 V of bias to the rotating substrate table. These two predeposition plasma processing steps are especially important to secure coating adhesion and stability at the coating-substrate interface. The rotation speed of the substrate turntable used in these LAFAD plasma processing trials ranged from 9 to 16 rpm. The substrates to be coated can be installed on a single rotation (SR) or a double rotation (DR) station located along the outer perimeter of the turntable. In SR mode, the substrates are rotated around the axis of the turntable with their front surface facing the chamber walls. In DR mode, substrates are subjected to double rotation around satellite station axes, as shown schematically in Fig. 1.

After the metal ion-etching step, the LAFAD coating deposition starts. During deposition of ceramic or cermet coatings, an ultrathin metallic bond layer is deposited first to further improve the coating adhesion. The metal coatings are deposited in argon, while during deposition of carbides, nitrides, or oxiceramic layers, a reactive gas (methane, nitrogen, and/or oxygen) is added to the processing chamber, forming a reactive gaseous plasma atmosphere. The gas pressure in LAFAD processes typically ranges from 0.2 to 1 mTorr for the deposition of ceramic coatings, while it can be as low as 0.01 mTorr or less for the deposition of diamond-like carbon (DLC) and related coatings.^{1,11} Note that during deposition of near dielectric oxiceramic and oxinitride coatings, a 13.56 MHz rf generator was used as a substrate bias power supply, while for the deposition of conductive coatings, an MDX-II (Advanced Energy) power supply, coupled with their Sparcle-V accessory, was used as a dc bias power supply. The substrate bias during deposition of most of the conductive coatings reported in this work was -40 V, while during deposition of oxiceramic and oxinitride coatings, the autopolariation bias potential ranged from -40 to -100 V.

Coating thickness was determined by the CALO™ wear scar spherical abrasion technique and optical micrometry to an accuracy of $\pm 0.1 \mu\text{m}$. Coating hardness and Young's modulus were measured by an MTS-XP nanoindenter with a continuous stiffness measurement (CSM) module and Berkovich tip. The CSM module allows for continuous measurement of hardness and modulus as a function of depth in the coating. In this approach, about 150 hardness and Young's modulus data points for each indent were averaged over an indentation depth range of approximately 5%–10% of a given coating thickness to improve the accuracy of measurement and to minimize any substrate effects; in addition, ten indent locations per sample were averaged to obtain statistical results. Adhesion evaluation was performed by the Rockwell C 145 kgf indentation test according to Ref. 13. Coating residual stress was calculated using the Stony method with coatings deposited on silicon beams. Average coating rms roughness was measured per ASTM B46.1 with a Veeco Dektak8 contact profilometer. Parameters for rms measurement were as follows: 5 μm radius stylus, 1750 μm scan length, 250 μm cutoff filter (waviness filter), and 1 data point/ μm . Five rms scans were made per sample and a minimum of six samples were scanned resulting in a 30 point rms average per unique coating type reported. Metallurgical cross sections were prepared for selected coatings, followed by scanning electron microscopy (SEM) microimaging. Coating compositions were analyzed by energy dispersive spectroscopy (EDS), Rutherford backscattering spectroscopy (RBS), x-ray photoemission spectroscopy (XPS), and Auger techniques.

III. RESULTS AND DISCUSSION

A. LAFAD vapor plasma transport characterization

The variation in the magnetic field strength along the axis of the primary cathodic arc source (axis “x” in Fig. 1) starting from the front surface of the primary DCAD targets in the LAFAD plasma source is shown in Fig. 3(a). It can be seen that the x -component of the cusp-type LAFAD magnetic field diminishes and reaches zero at the anode separator or dividing baffle at the center of the plasma guide chamber, where the magnetic field starts to turn 90° toward the coating chamber. On the other hand, the B_z component of the LAFAD deflecting magnetic field strength is low near the primary cathodic arc targets and increases to about 100 G inside of the plasma guide chamber and in the exit tunnel portion (see Fig. 1). The B_z component of the deflection magnetic field changes its sign inside of the plasma guide chamber at the $B=0$ area, as shown in Fig. 3(b), which is commonly referred to as the magnetic cusp node.^{8,14} The focusing and deflecting longitudinal magnetic field, which is aligned along the arc plasma flow, plays a crucial role in metal vapor plasma transport from the arc target down the tubular anode, around the corner at the entrance of the plasma guide tunnel, and finally out through the plasma guide exit tunnel and into the coating chamber.^{7,8} The elec-

trical conductivity across the magnetically insulated arc plasma column is reduced, resulting in an increased arc voltage drop in the filtered arc columns.

In the FCAD process, the focusing and deflecting magnetic fields are large enough to control the motion of the electrons but are not high enough to significantly affect heavy metal ions, as described by the following inequalities:^{1,3,4}

$$\rho_{Le} \ll L_p \ll \rho_{Li}, \quad (1)$$

where L_p is the characteristic size of the plasma duct, and ρ_{Li} and ρ_{Le} are the Larmor radii of ions and electrons, respectively, in the filtered arc plasma. When the electrons in the arc plasma flow become trapped on magnetic field lines, their diffusion across the filtered arc plasma column is suppressed, creating an excessive negative charge in the arc flow which attracts and confines the ions to secure the quasineutrality of the metal vapor plasma.^{1–4}

While the distribution of particles and their respective currents in a straight plasma flow, confined in a longitudinal magnetic field, does not depend on the polarity of the magnetic field streamlines, this is not the case for curvilinear filtered arc plasma flow. The direction of the magnetic field, which is determined by the polarity of the coils along the path of the filtered arc plasma from the primary DCAD source toward coating chamber, affects the position of the arc column. This is due to the shift of the arc column in a direction transverse to the plane of rotation of the filtered arc plasma flow moving along the curvilinear deflecting magnetic field from the cathode target along the anode of the primary DCAD source and further around the corner of the plasma duct exit tunnel toward the coating chamber. This effect is explained by the centrifugal drift of magnetized electrons moving with finite velocity along their trajectory in a curvilinear magnetic field.⁴ When the leading centers of the magnetized electrons' Larmor circles move along a curvilinear magnetic force tube having magnetic field $\mathbf{B}$ and radius of curvature $\mathbf{R}$ with velocity $v_{e\parallel}$, they will drift in a direction transverse to the plane of the magnetic force tube with the following centrifugal drift velocity v_{eCD} :¹⁴

$$v_{eCD} = (m_e c / q_e) |v_{e\parallel} / R|^2 [\mathbf{R} \times \mathbf{B}] / B^2, \quad (2)$$

where m_e and q_e are the mass and charge of the electron, and c is the speed of light. Due to the vector cross product in Eq. (2), it follows that the drift velocity will be in a direction perpendicular to the plane of the curving magnetic field line; furthermore, the direction of the drift will reverse if the polarity of $\mathbf{B}$ is reversed. Hence the actual path of the arc plasma flow will not be a strictly two-dimensional, planar curve; instead it will follow a three-dimensional, low-pitch helical path, as illustrated in Fig. 4. The sense of the helix depends on the polarity of the $\mathbf{B}$ vector as required by the $\mathbf{R} \times \mathbf{B}$ relationship. The direction of the drift velocity of magnetized electrons in the curvilinear filtered arc plasma flow is reversed when the polarity of the longitudinal deflection magnetic field is reversed. The positive metal vapor ions in a filtered arc flow follow the drift direction of the electrons

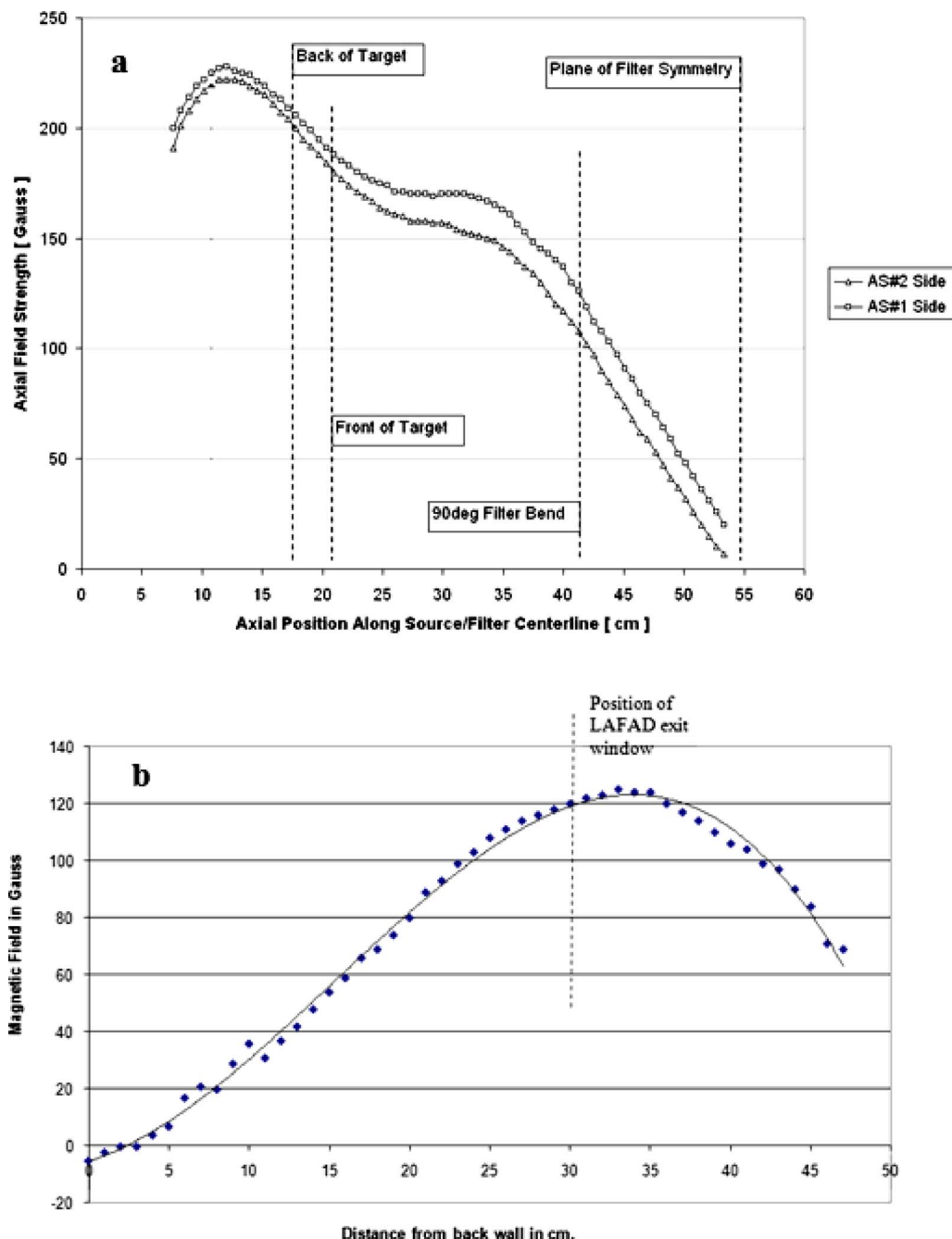


FIG. 3. (Color online) Intensity of the focusing and deflecting magnetic field in a primary DCAD source area and into the plasma guide chamber: (a) B_x component of the magnetic field along the path of a curvilinear focusing and deflecting longitudinal magnetic force line starting from the primary cathode target and going downstream along the plasma duct; (b) B_z component down the centerline of the plasma duct.

to satisfy the plasma quasineutrality requirement. Therefore, reversing the polarity of the magnetic focusing and deflecting longitudinal magnetic field along the filtered arc plasma flows will shift the arc plasma column and, accordingly, the distribution of ion and mass flow currents at the exit of the filter tunnel in a direction perpendicular to the plane of curvilinear arc flow, as illustrated in Fig. 4.

In our deposition system, the position of the primary cathodic arc sources attached to the opposite walls of the LAFAD plasma guide chamber are offset vertically from one another by 0.05 m to compensate for the effects of centrifugal

drift. The position of the primary cathodic arc source targets results in a shift of the positions of the arc spots on the evaporating surfaces of the cathode targets and, accordingly, the initial displacement of two opposite filtered arc columns in relation to each other. In this case, by changing the direction of the longitudinal curvilinear focusing and deflecting magnetic fields applied along the filtered arc flows, the arc columns can be shifted in opposite directions, either placing them close to each other or increasing the distance between them in the vertical direction, perpendicular to the plane of rotation of the filtered arc flow in the LAFAD

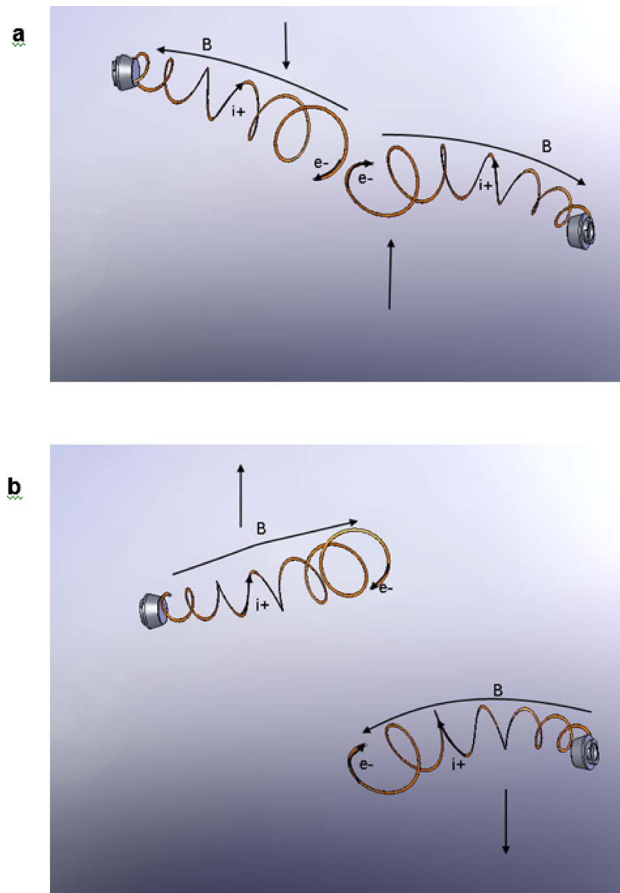


FIG. 4. (Color online) Schematic illustration (not to scale) of the cyclotron trajectories of the magnetized electrons along the curvilinear focusing and deflecting magnetic fields of the LAFAD source starting from the left and right primary cathode targets. Note the shift of the electron cycloids in opposite vertical directions depending on the direction of the longitudinal magnetic field: (a) filtered arc flows when the deflecting magnetic field is in the “in-plane” direction; (b) filtered arc flows when the deflecting magnetic field is in the “off-plane” direction.

plasma guide chamber, as shown schematically in Fig. 4. The direction of the longitudinal magnetic field collinear to the filtered arc plasma flow, which tends to force the two arc plasma columns to separate from each other, can be designated as the “off-plane” magnetic field. The reversed direction of the longitudinal magnetic field, which effectively attracts two curvilinear filtered arc plasma columns and makes them drift toward each other, can be designated as the “in-plane” magnetic field.

The ion current and mass flow distributions across the filter window were measured for two opposite polarities of the longitudinal magnetic field. These measurements were taken separately for the primary DCAD arc source 1 and the opposite primary arc source 2. The results of these measurements, shown in Fig. 5, demonstrate that the center (maximum intensity point) coincides both for ion current and for mass flow distribution in the vertical direction parallel to the axis of the vacuum chamber. Reversing the polarity of the longitudinal magnetic field shifts the centers of the filtered arc columns generated by arc sources 1 and 2 in opposite

directions. In one polarity the centers of ion current distribution in a filtered arc plasma flows generated by the primary sources 1 and 2 are shifted away from each other in the vertical direction, while in a reverse polarity they are moved closer to each other due to centrifugal drift effect in agreement with expression (2).

Figure 6 shows the comparison between vertical ion current distributions in carbon and titanium filtered arc flows measured by the vertically sectioned collector probe installed at the edge of the turntable in the coating chamber in front of the LAFAD plasma source exit tunnel opening. It can be seen that the positions of the center in both distributions coincide, but the absolute value of ion current is higher in titanium plasma than in carbon plasma, which can be explained by the large quantity of multicharged ions in a titanium vacuum arc, whereas mostly C^+ ions are present in a carbon arc.^{1,5,6}

The lateral distributions of the ion current in the filtered arc flows generated by the primary arc source 1 were measured by the laterally sectioned collector probe installed at the edge of the turntable in the coating chamber in front of the LAFAD plasma source exit tunnel opening with and without the short dividing anode baffle mounted in the plasma guide chamber between the tunnel entrance and the back wall. This is shown schematically in Fig. 1 (end positions for the short dividing baffle). The results of these measurements, shown in Fig. 7, demonstrate that when the dividing baffle does not cut the path of the filtered arc plasma flow at the turning point between the source anode and the filter exit tunnel, the lateral ion current distribution is not changed and the ion current value is reduced only slightly. The effect of the dividing baffle on the emerging filtered arc plasma mass flow and associated deposition rate was further investigated by the deposition of Ti/TiN and TiCr/TiCrN multilayer coatings on rotating substrates installed at the same location on the edge of turntable as that of the sectioned collection probe. The total primary cathodic arc current per each primary titanium arc target during deposition of Ti/TiN coatings was 140 A, which includes the primary arc current of 120 A and the auxiliary arc current of 20 A. In these processes, Ti sublayers were deposited in argon, while TiN sublayers were deposited using nitrogen as a reactive gas. In the case of TiCr/TiCrN coatings, the primary DCAD source 1 was equipped with a Ti target, while the opposite DCAD source 2 was equipped with a Cr target. The total arc current of the Cr cathode target was 170 A, while the total current of the Ti target was held at 140 A. The results of these trials, presented in Fig. 8(a), have demonstrated that the deposition rate of multilayer coatings deposited with a dividing baffle, which is short enough not to cut the turning filtered arc plasma flow, did not considerably differ from the deposition rate without the dividing baffle. However, on the other hand, when the dividing baffle extends into the exit tunnel and partially cuts off the path of the filtered arc plasma transport (medium length baffle position m in Fig. 1), then it substantially reduces the emerging vapor plasma flow which can be seen from the nearly factor of 2 reduction in

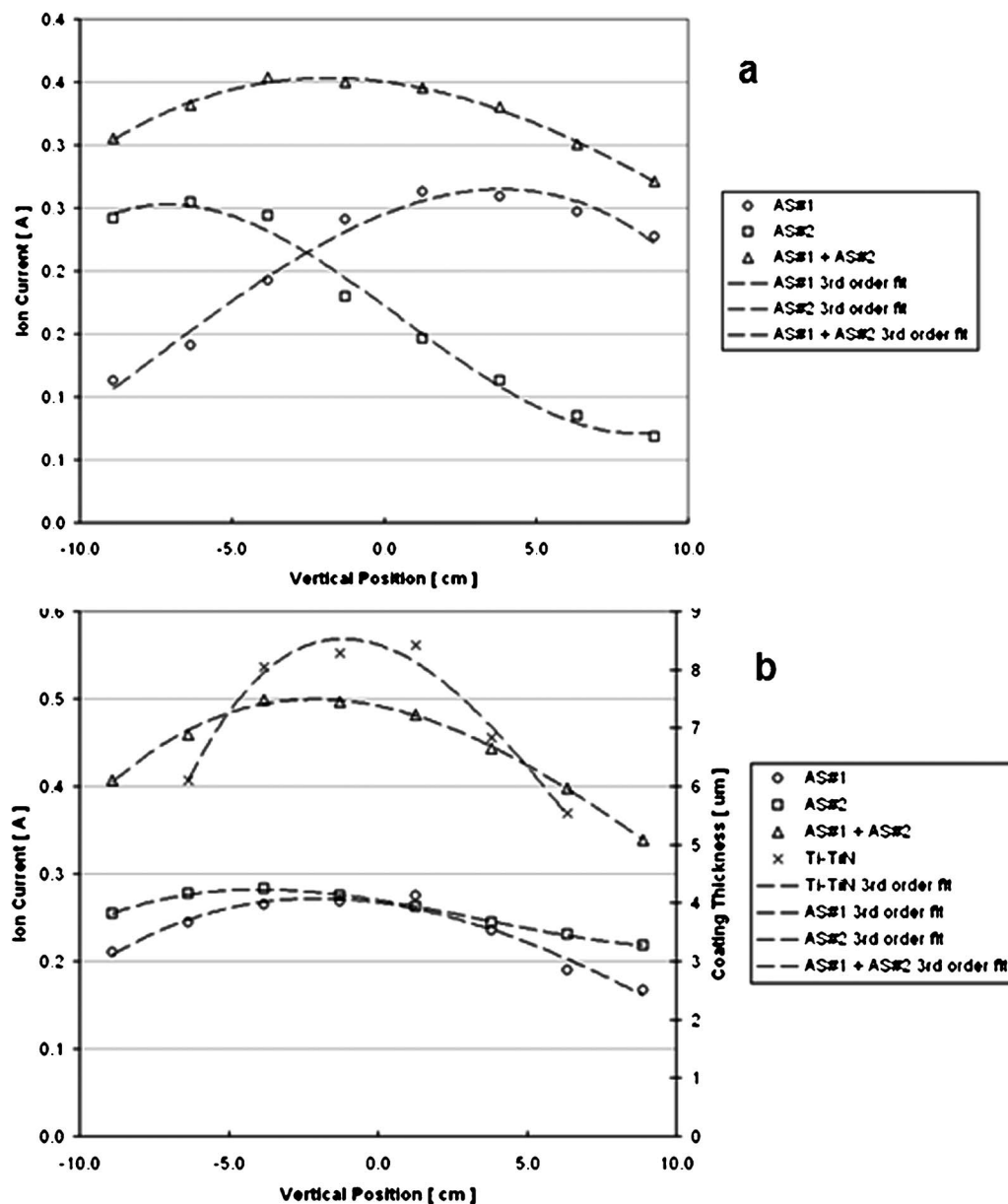


FIG. 5. Ion current and mass flow (deposition rate) distributions in the coating chamber in front of the exit window of the LAFAD source as a function of vertical position when the deflecting and focusing magnetic fields were oriented (a) in the off-plane direction; (b) in the in-plane direction. Note: ion current measurements were made for each source operated independently and for both sources operating together.

the deposition rate, as shown in Fig. 8(a). The deposition rate of Ti/TiN multilayer coatings deposited on substrates installed on the rotating turntable at the double rotated stations is 60% less than that of coatings deposited on single rotating substrates installed at the edge of the turntable. When the dividing baffle extends to the exit of the tunnel and completely isolates the left and right filtered arc flows in the LAFAD process, the deposition rate of the Ti/TiN coatings deposited on substrates installed on double rotating stations of the turntable is decreased by almost a factor of 3 compared to the same coatings deposited with the short dividing baffle.

The overall reduction in the deposition rate of TiCr/TiCrN coatings compared to TiN base composition can be attributed

to the much higher sputtering rate for multicharged Cr ions in comparison to Ti ions in a fully ionized LAFAD metal vapor plasma flow.¹⁵ This is illustrated in Fig. 8(b), which shows the dependence of Ti/TiN and TiCr/TiCrN multilayer coating deposition rates on the substrate bias. The reduced deposition rate of chromium-containing coatings is also due to the reduced evaporation rate of chromium targets versus titanium targets in the vacuum arc process,¹ which can be also seen from Table I.

The total ion current in the filtered arc plasma flow generated by one primary DCAD source 1 at lower and higher pressures is shown in Fig. 9 as a function of primary cathodic arc current. The increase in ion current output with an increase in the gas pressure can be attributed to the gaseous

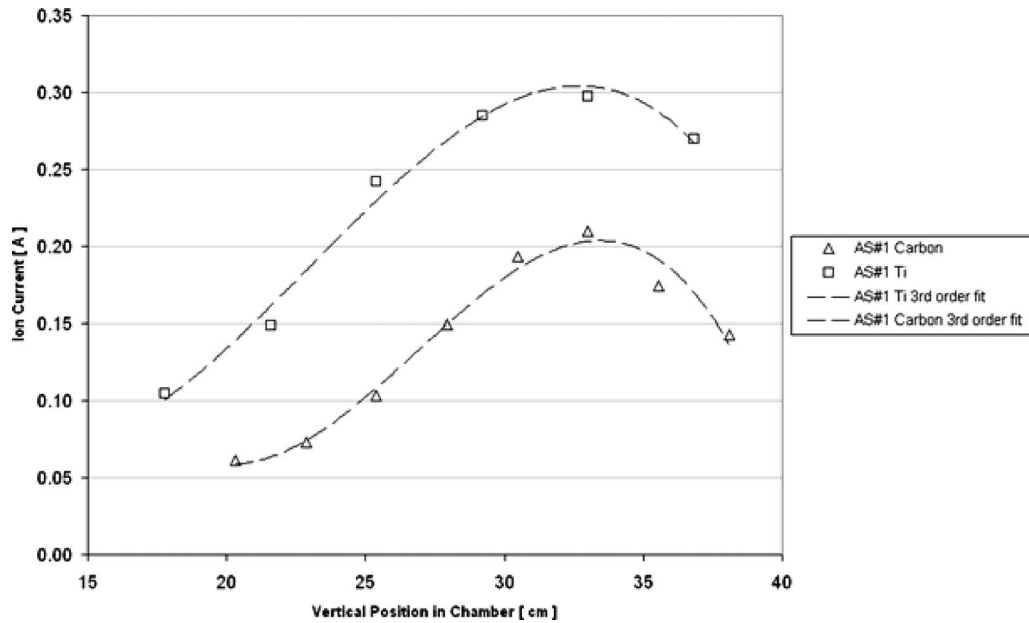


FIG. 6. Ion current dependence on cathode type and vertical position in the deposition chamber for the off-plane polarity magnetic deflecting/focusing field condition.

plasma component contribution. To make a conservative estimate of gaseous plasma component ionization rate in a simple case of Ti filtered arc flow with Ar as a buffer gas, one can assume that the increase in ion current with an increase in the argon pressure can be totally attributed to the contribution of the gaseous plasma component. In this case the average atomic flow density of argon ions across the filter exit tunnel can be estimated as follows:

$$w_{Ar^+} = [I(0.3 \text{ mTorr}) - I(0.06 \text{ mTorr})] / q_e A_{\text{filter exit}}, \quad (3)$$

where $I(0.06 \text{ mTorr})$ and $I(0.3 \text{ mTorr})$ are the total ion current collected at the exit of the LAFAD plasma source at argon pressures of 0.06 and 0.3 mTorr, respectively, $q_e = 1.6 \times 10^{-19} \text{ C}$ is the electron charge, and $A_{\text{filter exit}} = 0.12 \text{ m}^2$ is the area of the LAFAD source exit tunnel opening. The density of argon atoms, which cross the exit opening of the

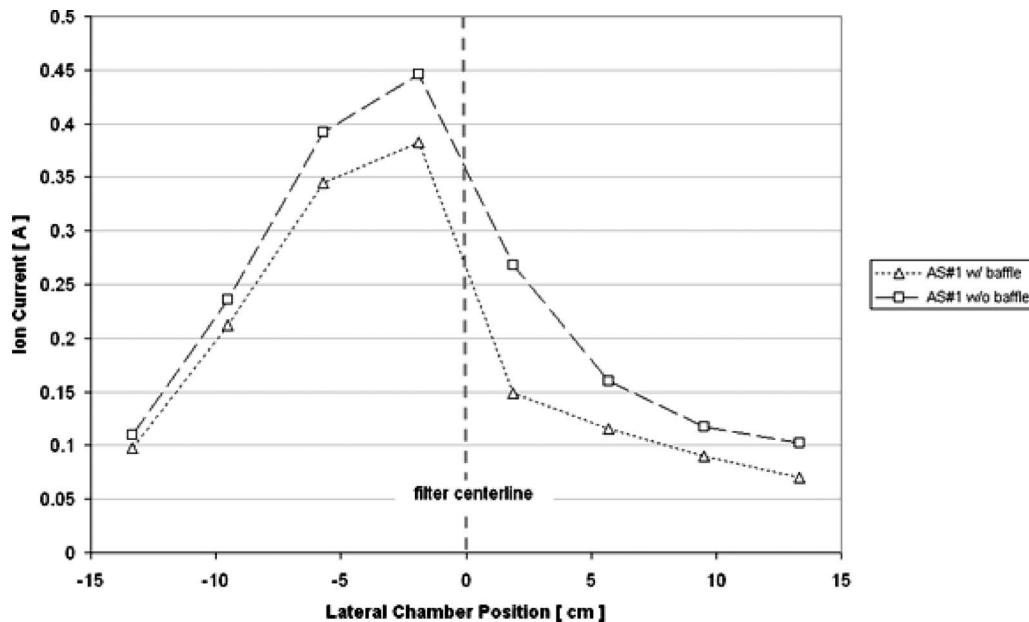


FIG. 7. Ion current as a function of the lateral position at the filter plasma duct exit. Two cases are shown; one with the dividing anode baffle installed in the filter duct, and the other without the dividing anode baffle installed. It is apparent that plasma efficiency is reduced when the dividing anode baffle is installed. Ti cathode at 100A, Ar at $5 \times 10^{-5} \text{ Torr}$, -36 V collector bias, AS 1 position is left of the filter centerline.

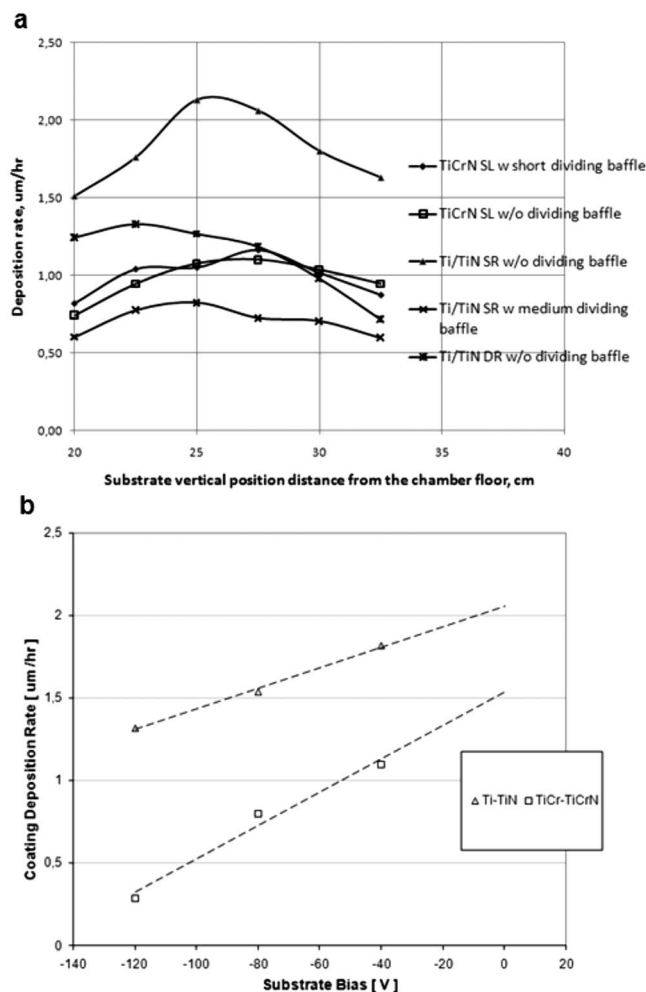


FIG. 8. Vertical uniformity and deposition rates of different LAFAD coatings deposited with and without dividing baffles (a) and at different substrate bias values (b).

LAFAD source, is given by the kinetic theory of gases¹⁶

$$w_{Ar} = 3573 \times 10^{26} P_{Ar} / (M_{Ar} T_{Ar})^{1/2} \text{ s}^{-1} \text{ m}^{-2}, \quad (4)$$

where P_{Ar} is the argon pressure in Torr, M_{Ar} is the molecular mass of argon, and T_{Ar} (K) is the temperature of a neutral atom gas in the chamber, which, for the sake of simplicity, can be considered ambient temperature. The ionization rate of argon in filtered arc plasma can be expressed as follows:

$$k_i(\text{Ar}) = w_{Ar^+} / (w_{Ar} + w_{Ar^+}). \quad (5)$$

Using Eqs. (3)–(5) and the data presented in Fig. 9, the argon ionization rate in the filtered arc plasma at the exit of the LAFAD plasma source can be estimated to be greater than 50%. It can be seen from Fig. 9 that the average argon ionization rate increases when the arc current increases, which can be explained by an increase in the intensity of argon ionization by electron-argon atom collisions when the arc current in filtered arc column increases. This also indicates that the area where most of argon ionization happens is the area inside of the arc plasma column and, more specifically, the current transfer area where the plasma density reaches its maximum value.^{1,4,12,17,18} The electron temperature in the vacuum arc column in the longitudinal external magnetic field typically ranges from 1 to 5 eV.¹ Assuming local thermodynamic equilibrium in the arc plasma column, the thermodynamic calculation of argon ionization rate was provided using TERRA (formerly named ASTRA) thermodynamic equilibrium software in the same way that it was used in Ref. 18. The results of this calculation, presented in Fig. 10, have shown that at temperatures above 1 eV (11 000 °C) and in the pressure range from 0.1 to 100 mTorr, which corresponds to the conduction area of the filtered arc column, the Ti–Ar vacuum arc plasma is fully ionized. The electron density increases when the pressure increases. The Ar^+ ion concentration drops when the Ti vapor concentration in the arc column increases. These results are in good agreement with the experimental measurements of ion concentrations in magnetically insulated vacuum arc columns.^{4,17}

The deposition rate in the LAFAD process also depends on the target material and the condition of the target surface as in any other processes using vacuum arcs.¹ The evaporation rate of titanium targets in a vacuum arc process exceeds the chromium evaporation rate by almost a factor of 2, as shown in Table I. The evaporation rate of used targets drops by 20%–25% depending on the usage time and arc current. This can be explained by a reduction in the target evaporation rate when the target is eroded; its front evaporation surface, which gets all the energy influx conveyed from the arc plasma, moves backward and the distance between the water-cooled back of the target and the front surface of the target (where the evaporation is actually taking place) diminishes. This results in dramatic decrease in temperature of the front evaporation surface of the cathode target, which, in turn, leads to a decrease in evaporation rates. It was shown that in the case of titanium targets the evaporation rate decreases by a factor of 2 when the temperature of the front evaporation surface of the cathode target decreases from 800 to 390 K.¹⁹ This effect is demonstrated in Fig. 11 which shows the deposition rates of TiN coatings deposited from two Ti primary arc sources with used and new cathode targets onto substrates installed on single rotation substrate holders of the LAFAD-500C turntable. It can be also seen that vertical rastering substantially increases the vertical coating thickness uniformity in the LAFAD process, increasing the $\pm 15\%$ coating thickness uniformity deposition area in the vertical direction from 15 to 25 cm.

TABLE I. Primary DCAD source target usage.

Left/right primary DCAD targets	Left/right targets total current (A)	Left/right targets erosion rate (mg/C)
Ti(new)/Ti(new)	140/140	63.9/65.1
Ti(new)/Ti(new)	200/200	102.6/103.5
Ti(used)/Ti(used)	200/200	76.3/77.5
Cr(used)/Cr(used)	200/200	29.2/27.8
Ti(used)/Cr(used)	200/200	69.7/30.0
TiCrAlY (new)/CoMn(no data)	120/100	102.1/(no data)

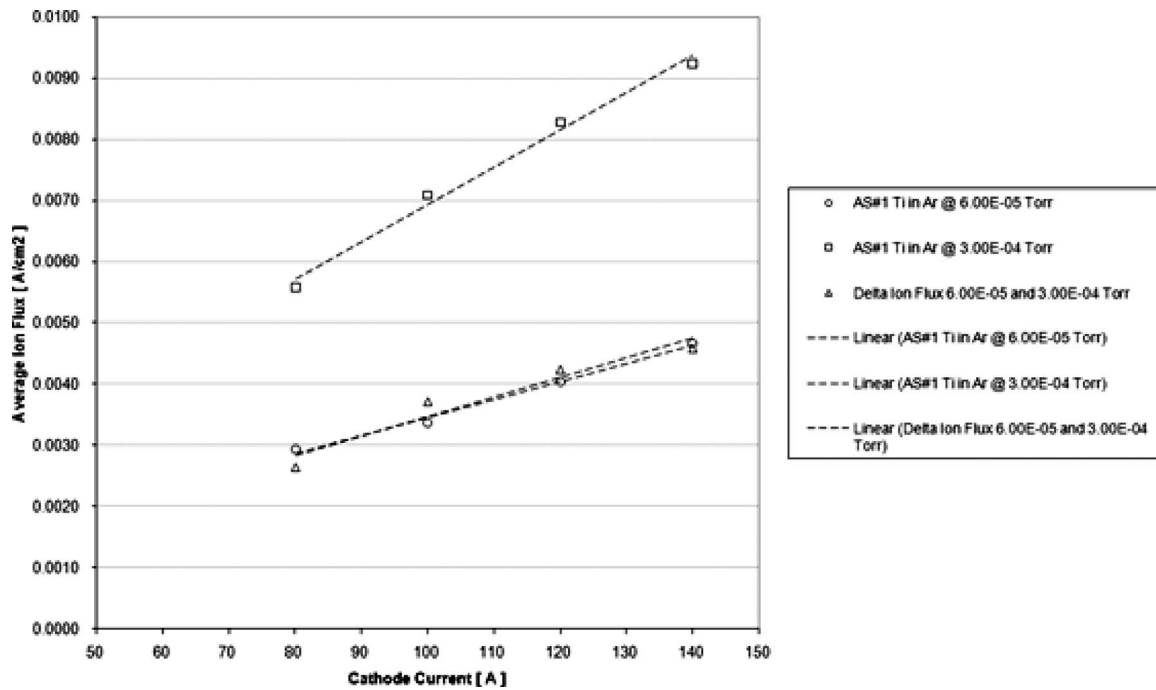


FIG. 9. DCAD source 1 total ion current as a function of cathode current and argon gas pressure, measured with DCAD source 2 turned off.

B. LAFAD coating characterization

A large variety of LAFAD coatings of different metal, ceramic, and cermet compositions and architectures depos-

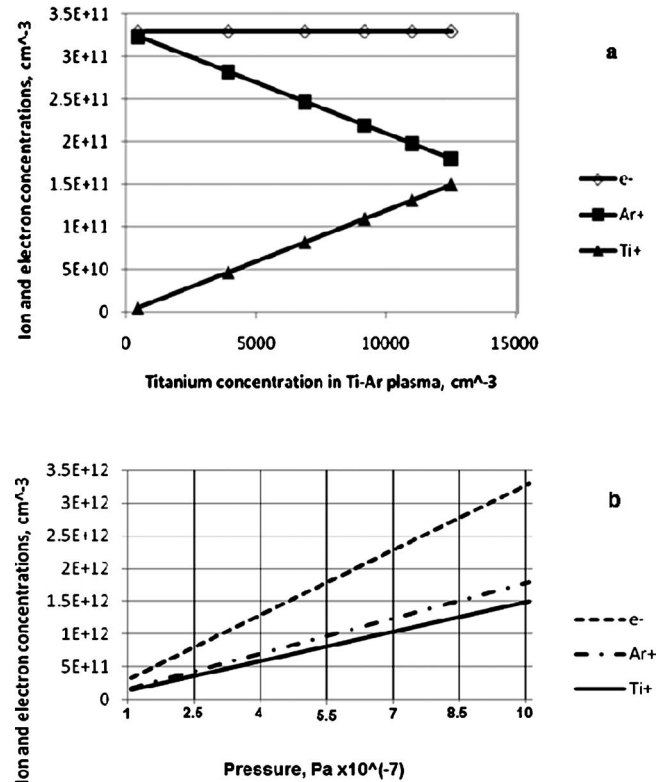


FIG. 10. Thermodynamic calculation of the ion and electron concentrations in a Ti-Ar plasma, assuming local equilibrium. (a) Concentration of charged particles vs concentration of titanium at 1 eV electron temperature. (b) Concentration of charged particles vs pressure at 1 eV electron temperature.

ited on different substrates have been characterized. An overview of the basic properties and capabilities of LAFAD technology for some common and uncommon coatings is presented in this section. Single layer TiN and AlN coatings deposited by the LAFAD process have a similar deposition rates ranging from 2 to 2.5 $\mu\text{m/h}$ when the primary arc source currents do not exceed 140 A. Both coatings show fine grain columnar morphology and extremely low density of isolated defects. No nodules, large grains, voids, porosity, and other growth defects, which are common in magnetron sputtered coatings, are detected in the LAFAD process. TiN coatings deposited by the LAFAD process have stoichiometric Ti50:N50 composition and hardness ranging from 30 to 35 GPa with an elastic modulus greater than 450 GPa, as shown in Table II. A SEM microimage of a fractured 5 μm thick AlN coating deposited during a 2 h deposition time on WC/6%Co carbide substrates installed on the rotating turntable is shown in Fig. 12. This coating features a fine grain, polycrystalline structure with substantially submicron grain size without any influence of the initial substrate texture on the coated surface profile. The composition of this coating was analyzed by XPS and by deuterium base RBS technique, which is capable of distinguishing between nitrogen and oxygen atoms. It revealed a nearly stoichiometric Al(50)N(50) composition for this coating. This hard ceramic coating shows about 28 GPa hardness with a 377 GPa elastic modulus, exceeding the standard data values for bulk AlN ceramics. Four-point-probe measurements have demonstrated that the electrical resistivity in this coating is in excess of $10^{12} \Omega \text{ cm}$.

When the total arc current from each primary DCAD source of the dual arc LAFAD source is raised to 200 A, the deposition rate of TiN coating increases and can exceed

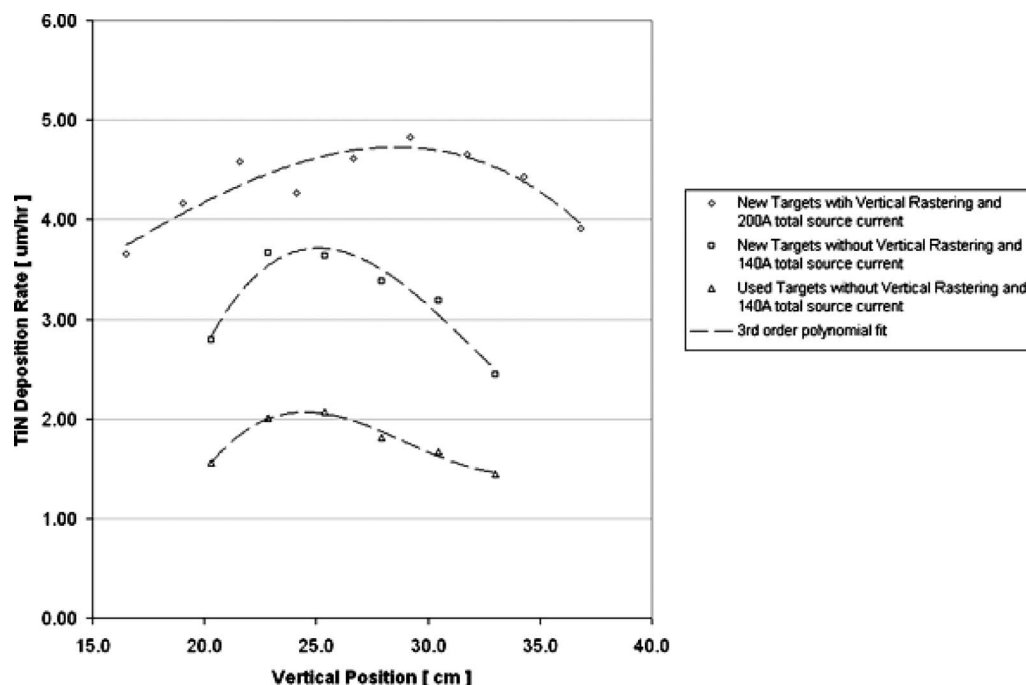


FIG. 11. Deposition rates of LAFAD Ti/TiN multilayer coatings deposited with and without vertical rastering field by used targets vs new targets installed on primary DCAD sources.

4 $\mu\text{m}/\text{h}$ even when vertical rastering is employed (Table II, item 3). Multilayer Ti/TiN coatings with relatively short bi-period and with thicknesses ranging from 3 to 5 μm (Table II, item 6) deposited in a batch coating system similar to that shown in Fig. 1 in single rotation and double rotation mode with vertical magnetic rastering are currently being used in the mass production of surface-engineered dental instruments.²⁰ Using a long dividing baffle, which effectively separates the filtered arc plasma flows generated by opposite primary DCAD sources 1 and 2 of the LAFAD source with different cathode targets, nanolaminated architectures with

nanometric size bi-period can be deposited, as was previously reported for AlCrN and TiCrN nanolaminated LAFAD coatings in Refs. 10 and 21.

Due to the high deposition rates of the LAFAD plasma source, thick Ti/TiN microlaminated cermet coatings were deposited on various substrates installed on single rotation and double rotation stations of the rotating turntable in the production batch coating setup of the LAFAD-500C coating system. Figure 13(a) demonstrates the extremely smooth, nearly defect-free surface of a 20 μm thick microlaminated Ti/TiN coating produced by one unidirectional LAFAD plasma source during a 10 h deposition period. The SEM image of the cross section of this coating, shown in Fig. 13(b), shows well-defined titanium and TiN sublayers with fine columnar structure. The 20 μm Ti/TiN coatings exhibit low residual compressive stresses, i.e., <1.5 GPa, resulting in exceptionally good adhesive and cohesive toughness, HF1 according to classification,¹³ as illustrated in Fig. 13(c). Low stresses in thick TiN base coatings may be explained by the role of thickness-dependent gradients of point defect density, as was recently proposed in Ref. 22. Optimization of Ti sublayer thickness versus TiN sublayer in this microlaminated cermet coating architecture plays an important role in energy dissipation by shear deformation, which delays critical shear and tensile stress developed at metallic/ceramic sublayer interface.²³ The hardness of these type of microlaminated coatings measured by nanoindentation technique exhibits a relatively low value in the range from 18 to 20 GPa, which can be considered as a composite hardness created by inter-

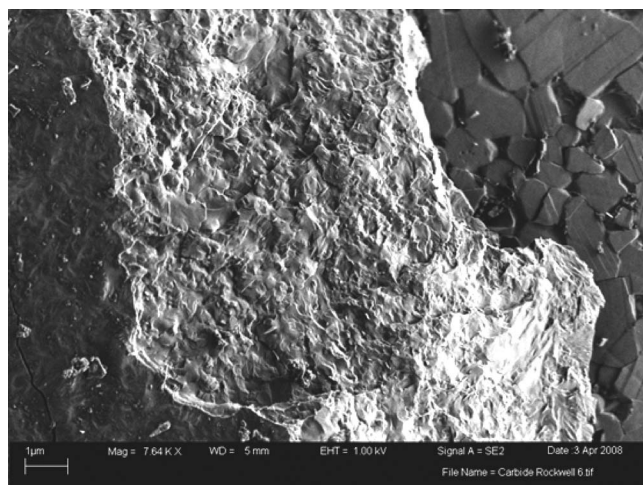


FIG. 12. SEM image of a fractured 5 μm thick AlN coating on a WC/6%Co carbide substrate.

TABLE II. Basic mechanical properties and composition of coatings deposited in the LAFAD-500C deposition system on 440A stainless steel substrates under various processing conditions. The substrates initially had a rms surface finish of 0.02 μm , and the substrates were mounted on the turntable in “single rotation” mode.

Item No.	Primary cathode target composition [total arc current per target I_{arc} (A)], left/right	Coating designation	Sublayers thickness ratio Me:MeX in bilayer, μm (Max sublayer deposition rate over 500 cm diameter $\times 15$ cm tall deposition zone, $\mu\text{m}/\text{h}$)	Average and (Max) rate over 500 cm diameter $\times 15$ cm tall deposition zone ($\mu\text{m}/\text{h}$)	Average and (Max) thickness over 500 cm diameter $\times 15$ cm tall deposition zone (μm)	rms roughness pre- (post deposition) (μm)	Hardness/elastic modulus (GPa)	Note
1	Ti(140)/Ti(140)	Ti-SL		0.7 (0.9)		0.02 (0.042)	9.5/197	
2	Ti(140)/Ti(140)	TiN-SL		3.19 (3.64)		0.02 (0.051)	35/484	
3	Ti(200)/Ti(200)	TiN-SL		4.83 (3.76)	7.65 (6.37)	0.02 (0.0824)	35/467	Fig. 11; deposition with vertical scanning Fig. 12
4	Al(120)/Al(120)	AlN-SL		(2.5)	(5)		28/377	
5	Ti20Cr30Al48Y2(120)/Ti20Cr30Al48Y2(120)	Ti11Cr26Al15Y1N42(O4)-SL		(6.96)	3.47 (2.99)		48/591	
6	Ti(140)/Ti(140)	Ti/TiN-ML	0.1/0.5 (2.04/2.66)	1.12 (1.33)	3.54(4.2)	0.02(0.23) 0.02(0.26)	28/394	
7	Ti(140)/Ti(140)	Ti/TiN-ML	0.68/1.77 (2.04/2.66)	(2.1)	(25.3)	0.02(0.23) 0.02(0.26)	19/301	Fig. 13
8	Ti(200)/Ti(200)	Ti/TiN-ML	(6.7/3)	4.67 (5.2)	(50)		18/295	Fig. 14
9	Cr(170)/Cr(170)	Cr/CrN-ML	1:5	0.38 (0.41)	1.5 (1.7)	0.02(0.078)	31/355	
10	Ti(140)/Cr(170)	TiCr/TiCrN-ML	1:2	1.0 (1.2)	2.0 (2.4)	0.02 (0.062)	32/405	
11	Ti(200)/Cr(200)	TiCr/TiCrN-ML	1:2	1.6 (1.8)	3.25 (3.68)	0.02 (0.058)	32/382	
12	Cr(200)/Cr(200)	CrC	N/A	1.3 (1.53)		0.02/0.072	28/389	
13	Ti20Cr30Al48Y2(140)/Ti20Cr30Al48Y2(140)	Ti5Cr19Al21Y1O54-SL	N/A	5.25 (6.3)	(12.6)	0.02/0.07	30/290	Figure 15
14	Ti20Cr30Al48Y2(140)/CoCrAlY(120)	Co9Ti4Cr18Al18Y1O49-SL	N/A	3.06 (3.8)	(4.24)	0.02/0.05	27/255	
15	Ti20Cr30Al48Y2(100)/Co50Mn50(120)	Co6Mn6Ti3Cr11Al18Y2O54-SL	N/A	3.94 (4.61)	(8.68)	0.02	21/243	Figure 16
16	Ti20Cr30Al48Y2(140)/Ti20Cr30Al48Y2(140)	Ti6Cr17Al23Y1O48N6-SL	N/A	4.79 (4.13)	(7.99)	0.02/0.06	27/265	

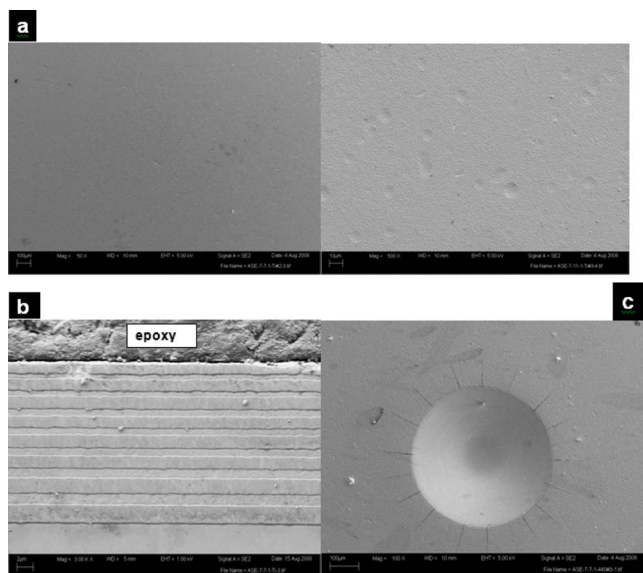


FIG. 13. (Color online) SEM images of a 20 μm thick Ti/TiN microlaminated LAFAD coating deposited during 11 h of deposition time from one LAFAD source on substrates installed on the rotating turntable of the batch coating system. (a) SEM images of coated surface at 50x (left) and 500x (right) magnification. (b) Metallurgical cross section. (c) Rockwell C 145 kg load indentation.

action of soft and ductile metallic sublayers with hard and brittle ceramic sublayers in Ti/TiN microlaminated coating architecture.

The total primary cathodic arc current of each primary DCAD source of the LAFAD plasma source during deposition of the 20 μm Ti/TiN microlaminated coatings was 140 A resulting in a deposition rate of ~ 2.2 $\mu\text{m}/\text{h}$ with a total coating thickness ranging from 20 to 25 μm for different metallic-to-ceramic ratios, as shown in Table II, item 7. It can be seen both from comparison of the deposition rates of the single layer pure metallic Ti and TiN coatings (Table II, items 1 and 2) as well as by analyzing Ti versus TiN sublayer deposition rates in Ti/TiN multilayer coatings, that the deposition rate of TiN is much higher than that of metallic Ti. The difference in the deposition rate between Ti and TiN coatings can be attributed to the fact that, during reactive deposition, atoms of nitrogen are incorporated into the growing film in addition to the titanium atoms coming from the filtered arc vapor flow. This effectively adds one nitrogen atom per each depositing Ti atom in a coating.

When the total arc currents of the primary arc sources are increased to 200 A, the deposition rate of the microlaminated Ti/TiN coating increases to 5 $\mu\text{m}/\text{h}$ (Table II, item 8). This coating, deposited during a 10 h deposition process, has a total thickness of 50 μm . The morphology of this coating is coarser than that of similar coatings deposited by a LAFAD process with lower primary arc currents, as illustrated in Fig. 14(a). It exhibits a larger columnar grain structure with sharper boundaries which can be explained according to the Movchan–Demchishin–Thornton–Messier zone diagram¹ as being caused by much higher substrate temperature during deposition of this coating due to higher ion currents gener-

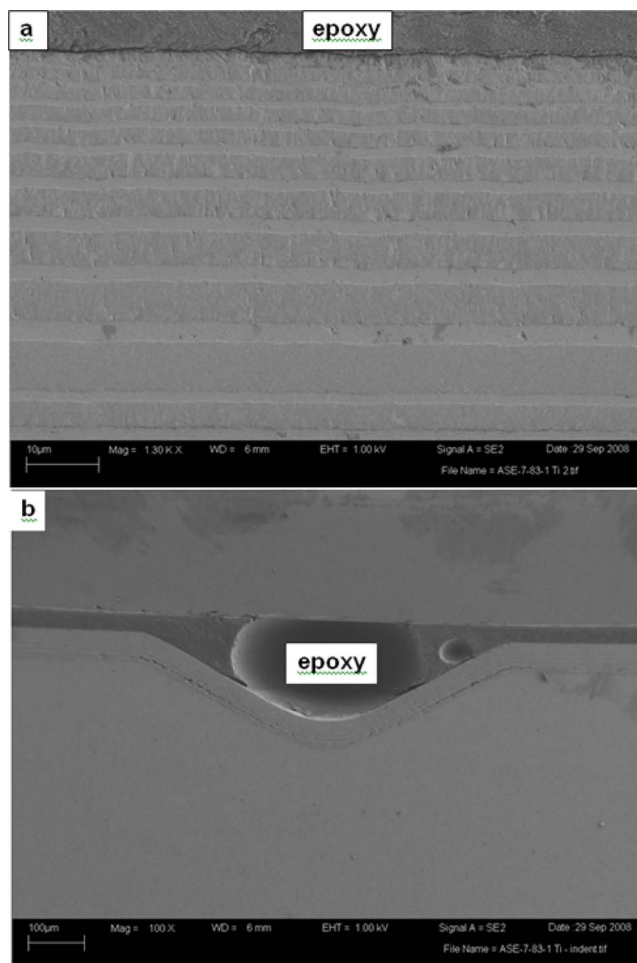


FIG. 14. (Color online) Cross section (a) and Rockwell C 145 kg load indentation mark (b) of a 50 μm thick Ti/TiN microlaminated coating deposited by one LAFAD source on rotating substrates in the batch coating system during a 10 h deposition time at 200 A total currents to the primary titanium DCAD sources.

ated by the primary arc sources of the LAFAD plasma source. It can also be seen that the bilayer period in this coating architecture changes from 10 μm at the beginning portion of the coating, interfacing the substrate, down to 3 μm at the top of the coating. This can be explained by a dramatic decrease in the evaporation temperature at the front surface of the targets and a corresponding reduction in evaporation rate, as targets are getting shorter via continuous evaporation as previously discussed in Sec. III A. This rate reduction can be compensated for via increasing the primary DCAD source current or adjusting the evaporation time per each coating bilayer period. The high adhesive and cohesive strengths of this thick microlaminated cermet coating are demonstrated by the Rockwell C 145 kg indentation mark shown in Fig. 14(b). It can be seen that only the bottommost thick ceramic TiN sublayer, interfacing the substrate, in this 50 μm thick microlaminated stack, which experience the largest mechanical deformation, exhibits the development of a larger lateral crack, while the network of a smaller micro-cracks develops mostly in the ceramic sublayers. These thick

microlaminated coatings are currently being used as corrosion and erosion protective coatings for turbomachinery components.²⁴

As already mentioned in Sec. III A, the deposition rate of the chromium base coatings is lower than that of Cr-free TiN base coatings due to higher resputtering rate of these coatings exposed to 100% ionized chromium vapor plasma,¹⁵ which consists of multicharged Cr ions, as was illustrated in Fig. 8(b), and also due to low evaporation rate of chromium primary arc targets, as was presented in Table I. The deposition rate of Cr/CrN coatings was found to be five times lower than that of Ti/TiN coatings, as shown in Table II (item 9). For the same reasons, the deposition rate of CrC single layer coatings is also less than that of TiN (Table II, item 14). The deposition rates of ternary TiCr/TiCrN base coatings are much higher than those of binary CrN coatings. The deposition rate of the ternary multilayer TiCr/TiCrN coatings depends on the Cr concentration in the Ti–Cr–N–Ar plasma and on the substrate bias. At low substrate bias, $V_b < -40$ V, the deposition rate of TiCrN base coatings approaches the rate of TiN base coatings. It can be seen that when bias voltage increases, the deposition rate of chromium-containing nitride coatings decreases. This effect is much less in the case of Cr-free TiN base coatings as was shown in Fig. 8(b). This may be explained by the lower resputtering rate of TiN base coatings by Ti metal vapor plasma in comparison to Cr-containing coatings exposed to highly ionized Cr-containing metal vapor plasma.¹⁵

The highest productivity of the LAFAD process was demonstrated during the deposition of AlCrO(N)-based multielemental oxiceramic and oxinitride coatings. This can be attributed to the influx of the oxygen and/or nitrogen atoms conveyed from the metal-gaseous plasma to the substrate surface. In the case of AlCrO-based coatings, the Me_2O_3 corundum phase prevails. Therefore per each depositing pair of Al or Cr atoms three oxygen atoms will be added from the reactive gas atmosphere, which will inflate the coating thickness even more than that of the previously discussed TiN coatings. In addition the primary cathode targets with a high concentration of aluminum have a greater evaporation rates than aluminum-free targets as can be seen from Table I. Figure 15 shows the EDS line scan of a 12 μm thick TiCrAlYO single layer coating deposited by the unidirectional LAFAD plasma source with two primary opposite DCAD sources 1 and 2 equipped with Ti20Cr30Al40Y2 targets on substrates installed on a rotating turntable of the batch coating system. The maximum deposition rate of this coating, deposited without vertical rastering of the plasma flows on substrates located near the center line of the LAFAD source exit window, exceeded 6 $\mu\text{m}/\text{h}$, while the average rate inside of the 0.15 m deposition area around the center line of the LAFAD source exit was 5.25 $\mu\text{m}/\text{h}$. Similar deposition rates have been achieved for a variety of oxiceramic, oxinitride, and nitride coatings based on TiCrAlY system with the addition of Co and Mn, as shown in Table II. It can be seen that the hardness of oxiceramic and oxinitride coatings of this elemental system ranges from 20 to 30 GPa, with superhard

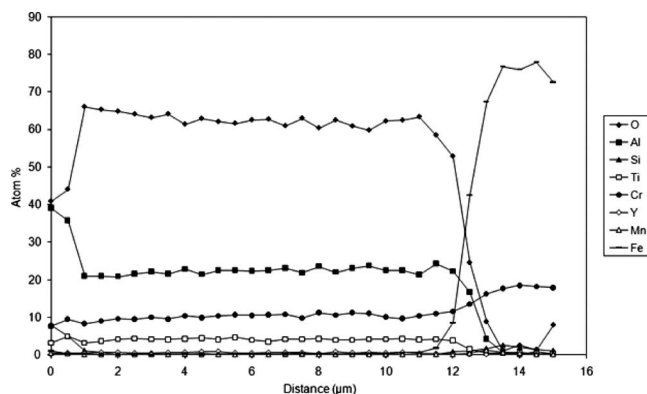


FIG. 15. EDS line scan of TiCrAlYO single layer coatings deposited by one LAFAD source with primary DCAD sources equipped with the same Ti20Cr30Al40Y2 targets during a 2 h coating deposition run on substrates installed on the rotating turntable of the industrial batch coating system.

properties demonstrated by the single layer TiCrAlYN nitride coating (see Table II, items 5 and 13–16). It was also found that the stress level in these coatings increases with increasing nitrogen content, correlating with the hardness. The assessment of electrical conductivity of these MTiCrAlY(O,N) coatings at ambient temperature using electric probe have demonstrated its near dielectric electrical resistivity as a result of defectless structure and the absence of inclusions of metallic particles.

The comparison of LAFAD TiCrAlY(O,N) coating compositions presented in Table II (items 5, 13, and 16) versus the primary DCAD targets elemental composition obtained by different surface analysis techniques are shown in Table III. RBS, d-RBS, XPS, EDS, and Auger electron spectroscopy (AES) techniques were used to measure the composition of the TiCrAlY primary DCAD source target material and coatings. Prior to AES analysis, the target material was given a brief argon ion sputter to remove surface contamination and oxidation. Since AES and EDS probe relatively small volumes, and the target is quite heterogeneous in composition, it is questionable whether this could be a useful standard for the analysis of the LAFAD coatings. The results in Table III indicate that both EDS and AES measure the composition of the TiCrAlY target to be chromium deficient and aluminum rich compared to the certified composition. This discrepancy is greatest with AES, which probes the smallest volume. AES is sensitive to only the top several nanometers, while EDS detects the composition within the top ~ 2 μm . AES analysis of homogeneous Al_2O_3 and TiO_2 standards gave compositions in good agreement with expected compositions. The Ti+Cr component concentration is also presented since RBS technique cannot reliably distinguish between Ti and Cr spectra. It can be seen that the results obtained by both XPS, EDS, and RBS analyses give a reasonably good agreement between metallic component concentrations in TiCrAlY(O,N) LAFAD coatings and the Ti20Cr30Al48Y2 primary DCAD targets, which demonstrates that LAFAD process is capable of depositing multi-

TABLE III. TiCrAlY target composition and metal elements concentrations in TiCrAlYN, TiCrAlYO, and TiCrAlYON LAFAD coatings as of Table II (items 5, 13, and 16), obtained by different surface analysis techniques.

Sample	Analysis	Ti	Cr	Ti+Cr	Al	Y
TiCrAlY target	Certified	20	30	50	48	2
TiCrAlY target	EDS	19.8	21.1	40.9	55.4	2.5
TiCrAlY target	AES	12.9	5.7	18.6	78.1	3.2
TiCrAlYO LAFAD coating	EDS	11.42	27.52	38.94	59.40	1.66
TiCrAlYO LAFAD coating	AES	9.96	41.42	51.38	46.46	2.15
TiCrAlYO LAFAD coating	RBS	N/A	N/A	40	59	1.6
TiCrAlYO LAFAD coating	XPS	8.5	19.5	28	64.2	7.7
TiCrAlYO LAFAD coating	d-RBS ^a	N/A	N/A	56	42	1.8
TiCrAlYON LAFAD coating	d-RBS	N/A	N/A	40	60	0.13
TiCrAlYN LAFAD coating	XPS	15.3	36.5	51.9	46.8	1.3
TiCrAlYN LAFAD coating	RBS	N/A	N/A	47.3	51.0	1.6
TiCrAlYN LAFAD coating	d-RBS ^a	N/A	N/A	56.4	41.8	1.9

^aRBS with deuterium beam.

elemental coatings having compositions that closely resemble the primary DCAD targets composition.

The EDS line scan in Fig. 16 shows the elemental distribution across a 3 μm thick CoMnTiCrAlYO coating deposited during a 40 min deposition time by the LAFAD plasma source with two different primary arc targets: a Ti20Cr30Al48Y2 target installed on the primary DCAD source 1 and a Co50Mn50 target installed on the primary arc source 2. It can be seen that the LAFAD process allows effective mixing of two filtered arc plasma flows generated by primary targets with different multielemental compositions to achieve deposition of the complex multielemental nanocomposite coatings at high deposition rates. The TiCrAlYO(N)-based coatings were initially developed as protective coatings for the interconnect plates of solid oxide fuel cells, but their potential range of applications is much wider, including thin film thermistors, protective coatings for hostile environments, high temperature erosion and corrosion resistant coatings for turbomachinery operating in hostile en-

vironments, coatings for cutting tools, especially for cutting superalloys, coatings for molds and dies, including die casting dies for alumina-silica foundries, and biocompatible coatings for medical and dental instruments and implants to name a few.^{7,9–11,20,21,24}

IV. CONCLUSIONS

The unidirectional dual arc LAFAD vapor plasma source has been characterized as a generator of high density ion current and mass flow. The intensity of the dual filtered arc flow generated by a LAFAD source increases when the deflecting and focusing magnetic fields along the filtered arc plasma streams are increased. The direction of the deflecting/focusing magnetic field affects the position of the filtered arc columns in the direction transverse to their plane of rotation in the LAFAD plasma guide chamber. The in-plane direction of the longitudinal magnetic fields allows the emerging filtered arc flows to stay close to each other. The uniformity of the distribution of ion current and mass flow at the exit of the LAFAD source and, accordingly, the uniformity of the coating deposition rate can be controlled by a superimposed vertical rastering magnetic field. The ionization rate of metal-gaseous filtered arc plasma produced by a LAFAD source is 100% for the metal vapor component and exceeds 50% for the gaseous plasma component. The productivity of the LAFAD plasma source is comparable to, or exceeds, the productivity of conventional DCAD sources and magnetron sputtering sources for a wide variety of metal, ceramic, and cermet coatings of different compositions and architectures tested and evaluated in this study. For example, one LAFAD source can deposit TiN-based coatings on substrates installed on the 0.5m diameter rotating turntable of the industrial size batch coating chamber with a productivity of up to 5 $\mu\text{m}/\text{h}$. In the case of AlCrO(N)-based oxiceramic and oxinitride coatings, the productivity increases to more than 6 $\mu\text{m}/\text{h}$ with a coating metallic component composition being nearly identical to the primary cathodic arc targets composition. These coatings are also characterized by nearly defect-free

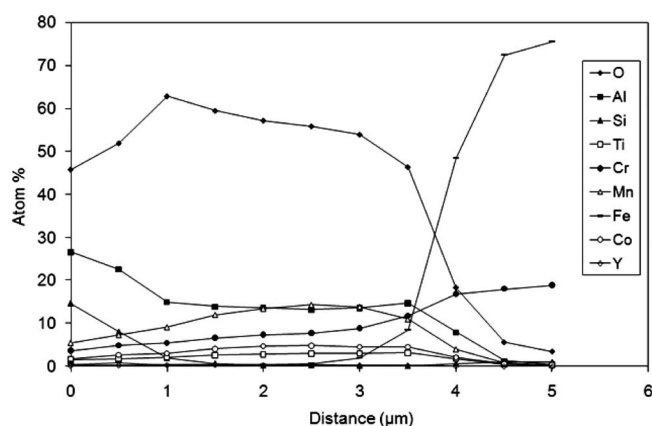


FIG. 16. EDS line scan of CoMnTiCrAlYO single layer coatings deposited by one LAFAD source with primary DCAD source 1 equipped with a Ti20Cr30Al40Y2 target and DCAD source 2 equipped with a Co50Mn50 target during a 40 min coating deposition run on substrates installed on the rotating turntable of the industrial batch coating system.

morphology, and an extremely smooth surface without the growth defects that are typical in DCAD and magnetron processes, such as nodules, large grain, voids, and porosity. The hardness of ceramic and cermet coatings deposited by the LAFAD process ranges from 20 to 40 GPa for multilayer nitride base and oxiceramic coating and up to a superhard level of nearly 50 GPa for multielemental single layer nitride coatings. These properties of the LAFAD process make it an attractive alternative to replace conventional plasma PVD processes for a wide range of applications.

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