Epitaxial TiN(001) wetting layer for growth of thin single-crystal Cu(001)

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Single-crystal Cu(001) layers, 4-1400 nm thick, were deposited on MgO(001) with and without a 2.5-nm-thick TiN(001) buffer layer. X-ray diffraction and reflection indicate that the TiN(001) surface suppresses Cu-dewetting, yielding a 4 x lower defect density and a 9 x smaller surface roughness than if grown on MgO(001) at 25 °C. In situ and low temperature electron transport measurements indicate that ultra-thin (4 nm) Cu(001) remains continuous and exhibits partial specular scattering at the Cu-vacuum boundary with a Fuchs-Sondheimer specularity parameter p = 0.6 ± 0.2, suggesting that the use of epitaxial wetting layers is a promising approach to create low-resistivity single-crystal Cu nanoelectronic interconnects. © 2011 American Institute of Physics. [doi:10.1063/1.3624773]

I. INTRODUCTION

The integration of metals with chemically incompatible ceramic materials in terms of interfacial reactions, contamination, or processing ambient is important for a number of technological applications including semiconductor interconnects, coated high-temperature superconducting tapes and heterogeneous catalysis. Various studies have examined the metal/ceramic interface and, in particular, the growth mode, morphology, crystal quality, and strain relaxation of the growing metal film as well as technologically important properties including wetting/adhesion, electromigration, electron scattering, and oxygen and metal interdiffusion.

The growth of large-grain or epitaxial single-crystal Cu wires has recently gained considerable interest since it has the potential to suppress the dramatic resistivity increase associated with grain boundary scattering in Cu interconnect wires with decreasing width. This may potentially be realized by barrier layers that facilitate epitaxial Cu growth as well as the wetting of Cu on the dielectrics.

The epitaxial growth of Cu(001) layers on MgO(001) surfaces has been demonstrated earlier. However, this system is characterized by a large lattice mismatch of ~14%, which leads to a high density of misfit dislocations at the Cu-MgO interface, and the nucleation of misoriented grains for growth above 200 °C. Moreover, the surface energy of 1.9 Jm⁻² for Cu (Ref. 17) is larger than that of 1.1 Jm⁻² for MgO(001) (Ref. 18), resulting in a Volmer-Weber growth mode with 3D Cu clusters. Thus, dewetting inhibits the growth of a thin continuous Cu layer on MgO. The wetting of Cu on TiN is expected to be better, due to the surface energy which is larger than that of MgO(001), and has been predicted to be 1.3 Jm⁻² for relaxed stoichiometric TiN(001), but is even larger for the very common N-deficient TiNₓ surface. In addition, TiN(001) forms thin adherent epitaxial layers on MgO(001) and therefore represents an ideal candidate as an epitaxial wetting layer for Cu. Further, TiN layers are used and/or proposed as the diffusion barrier layers and metal gate materials in Si and GaAs device metallization because of their high melting point, good thermal and chemical stability, high hardness, low diffusivity, and suitable workfunction.

In this article, we demonstrate that the growth of Cu(001) on TiN(001), in fact, leads to a higher crystal quality, smoother surfaces, and thinner continuous layers than if deposited on MgO(001). This is attributed to the larger surface energy of TiN than of MgO, especially if the TiN surface is N-deficient. In addition, the ultra-thin continuous Cu(001) single-crystal films in this study facilitate an accurate quantification of the surface scattering process, because grain boundary scattering is absent, the low density of point defects and dislocations renders defect scattering negligible, and the layer thickness, d = 4 nm, is considerably smaller than the bulk mean free path of 39 and 313 nm at room temperature and 77 K, respectively. The specularity parameter for electron scattering at the Cu-vacuum and Cu-air interfaces is found to be 0.6 ± 0.2 and 0 ± 0.1, respectively. The results suggest the use of epitaxial TiN as a promising adhesion surface to deposit thin (< 50 nm) single-crystal metal layers with a low defect density, and to create low-resistivity Cu nanoelectronic interconnects.

II. EXPERIMENTAL PROCEDURES

All of the layers were grown in a three-chamber ultrahigh vacuum dc magnetron sputtering system with a base pressure of <10⁻⁹ Torr. The polished MgO(001) substrates were cleaned with successive rinses in ultrasonic baths of tri-chloroethylene, acetone, isopropyl alcohol, and de-ionized water and thermally degassed at 800 °C in vacuum. The 5-cm-diameter Cu (99.99%) target was facing the substrate at a distance of 20 cm and depositions were performed in 2.5 ± 0.2 mTorr Ar (99.999%), at a constant power of 150 W yielding a deposition rate of 0.21 nm/s. Cu layers of thickness d = 4-1400 nm were deposited at 25 and 80 °C on MgO(001) and on TiN(001)/MgO(001), where the TiN(001) is a 2.5-nm-thick single-crystal TiN underlayer that is in situ

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deposited at 700 °C with a rate of 0.07 nm/s by reactive sputtering in 3.5 mTorr pure N2, prior to the Cu deposition. In addition, 120-nm-thick-Cu layers were deposited on MgO(001) and on 560-nm-thick-TiN(001)/MgO(001) substrates for thin film strain measurements using high resolution x-ray reciprocal space mapping. After the deposition, the layers were transported without breaking vacuum to the analysis chamber for in situ resistivity measurements with a linear four point probe operated at 0.001-100 mA. The samples were allowed to self-cool to room temperature and thermal equilibrium was considered to be reached when the measured resistivity asymptotically approached a constant value that varied less than 0.1% per 1 h. The samples were then removed from the vacuum system by transferring to a load lock that was vented with dry N2 and were next dropped into liquid N2 within 2 s after removal from the vacuum system to minimize surface oxidation, followed by four point probe measurements at 77 K in liquid N2. After removal from the liquid N2, another four point probe measurement in air at room temperature (∼25 °C) was performed. For this purpose, the samples were blown dry with dry N2 to minimize ice or water buildup on the Cu surface. The layer thicknesses were determined from the temperature dependence in the resistivity and were verified using Rutherford backscattering and x-ray reflectivity (XRR). Layer roughness were also quantified by XRR analyses, which are in agreement with our previous in situ scanning tunneling microscopy results for Cu(001) layers on MgO(001) substrates. The crystallinity, texture, and in-plane strain of Cu layers were quantified by x-ray diffraction (XRD) using a Panalytical X’pert PRO MPD system with a Cu Kα source, two-crystal Ge(220) two-bounce monochromator that provides monochromatic Cu Kα1 radiation with a 0.0068° divergence for ω-2θ and ω-rocking scans, a point focus x-ray lens (poly-capillary optics) that provides a quasi-parallel Cu Kα beam with a divergence of less than 0.3° for pole-figure texture measurements to minimize defocusing effects, and a PIXcel line detector for fast and high resolution reciprocal space mapping where the reflection angle between the sample surface and the detector is small (< 20°) to reduce the beam width. The reciprocal space maps were obtained over an angular range of ±0.8° in both Δω-2θ and Δω, with a step size in both directions of 0.02°.

III. RESULTS AND DISCUSSION

Figures 1(a)–1(d) show representative x-ray diffraction (XRD) results from Cu/MgO(001) and Cu/2.5-nm-TiN(001)/MgO(001) layers. For all samples, the only observable peaks in the ω-2θ scans from 30 to 90° 2θ are the MgO 002 and Cu 002, indicating well-developed 001 preferred orientation of Cu. The plot in Fig. 1(a) is a small portion of the ω-2θ scans, showing the Cu 002 peaks from 40-nm-thick Cu layers grown on TiN(001) at Tsub = 25 °C and on MgO(001) at Tsub = 80 and 25 °C. The peak maxima correspond to the measured out-of-plane Cu lattice constants of 0.3606, 0.3611, and 0.3612 nm, respectively. These values are slightly below the reported bulk Cu lattice constant of 0.3614 nm,27 indicating a residual biaxial tensile stress which is attributed to the misfit with TiN and MgO of (αCu - αCu)/αCu = 14.78% and (αMgO - αCu)/αMgO = 14.18%, respectively. In addition, the tensile stress may be partially attributed to differential thermal contraction during the cooling to room temperature after deposition, since the thermal expansion coefficient of the Cu film, αCu = 16 × 10⁻⁶ K⁻¹ (Ref. 28), is larger than of the MgO substrate, αMgO = 9 × 10⁻⁶ K⁻¹ (Ref. 28). The Cu 002 peak full-width-at-half-maximum (FWHM) increases from 0.46° for Cu/TiN with Tsub = 25 °C, to 0.71° for Cu/MgO with Tsub = 80 °C, to 0.81° for Cu/MgO with Tsub = 25 °C. This corresponds to a decreasing x-ray coherence length of 19, 13, and 11 nm, respectively, which is below the layer thickness of 40 nm and indicates increasing residual strain variations and/or crystal-line defects within the Cu layer,18 which are also responsible for the decreasing peak intensity in Fig. 1(a). The corresponding Cu 002 ω-rocking curves are plotted in Fig. 1(b) and exhibit a FWHM of 1.1, 3.8, and 4.7°, respectively. This confirms the considerably higher crystalline quality for the layer grown on TiN(001) and indicates an ~3.5 × and ~4 × higher

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grain alignment for Cu/TiN than for Cu/MgO with $T_s = 80$ and 25 °C, respectively. The wider rocking curve for Cu/MgO indicates a lower crystalline quality, however, this may also be due to the tilt of the Cu grains to relieve misfit strain, as previously reported for Cu(001)/MgO(001).15,16

XRD pole figure measurements for all samples, using constant 2θ values corresponding to Cu 002 and Cu 111 reflections at 50.42 and 43.32°, respectively, show only a single peak for Cu 002 at the origin (not shown) and fourfold symmetric peaks at a tilt, $\psi = 54.7^\circ$, and at polar angles, $\phi = 45, 135, 225,$ and 315° for Cu 111, as presented in the representative Fig. 1(c) for a Cu/TiN layer. The pole figure analyses confirm a cube-on-cube epitaxial relationship of the Cu layers with the MgO substrates Cu(001) and MgO(001) in the Cu lattice constant of 0.3614 nm, the in-plane obtained from the range of 0.31-0.36 for bulk polycrystalline copper.29,30

As shown in Figs. 1(e) and (f), respectively. The RSMs are plotted as iso-intensity contours in a plane of the reciprocal space with coordinates $k_x$ and $k_y$ which yields Poisson ratios and the TiN wetting layers: Cu(001) epitaxial relationship of the Cu layers with the MgO substrates Cu/TiN layer. The pole figure analyses confirm a cube-on-cube oblique diffraction geometry. The TiN 113 peak, shown in any lateral irregularities would cause peak broadening for such respectively, indicating a comparable surface roughness since from the RSM are $\theta$-scan direction is 0.4 and 0.3° for the Cu/TiN layer is 0.47°, which is 3 × smaller than 1.4° for the Cu/MgO layer, indicating an ~3 × smaller grain mosaic spread for growth on TiN(001) than on MgO(001). The FWHM for the Cu 113 peak along the $\omega$-scan direction for the Cu/TiN layer is 2.6 × smaller than the value of 3.7° for both Cu/MgO layers, confirming the higher crystalline quality and better alignment of Cu on TiN(001) than on MgO(001).

High resolution reciprocal space maps (HR RSMs) were measured around the Cu, TiN, and MgO 113 reflections, for 120-nm-thick Cu layers grown on 580-nm-thick TiN(001)/MgO(001) with $T_s = 25$ °C and on MgO(001) with $T_s = 80$ °C, as shown in Figs. 1(e) and (f), respectively. The RSMs are plotted as iso-intensity contours in a plane of the reciprocal space with coordinates $k_x$ and $k_y$ parallel and perpendicular to the sample surface, pointing along the [110] and [001] crystalline directions, respectively. The FWHM for the Cu 113 peak along the $\omega$-scan direction for the Cu/TiN layer is 0.47°, which is 3 × smaller than 1.4° for the Cu/MgO layer, indicating an ~3 × smaller grain mosaic spread for growth on TiN(001) than on MgO(001). The FWHM for the Cu 113 peak along the $\omega$-2θ scan direction is 0.4 and 0.3° for Cu/TiN and Cu/MgO, respectively, indicating a comparable surface roughness since any lateral irregularities would cause peak broadening for such an oblique diffraction geometry. The TiN 113 peak, shown in the inset in Fig. 3(e), has a FWHM of 0.04° and <0.01° along $\omega$ and $\omega$-2θ, respectively, indicating a strong alignment of the TiN(001) to the MgO(001) surface. The in-plane $a_{||} = \sqrt{2}/k_{||}$ and out-of-plane $a_{\perp} = \sqrt{2}/k_{\perp}$ Cu lattice constants obtained from the RSM are $a_{||} = 0.3622$ nm and $a_{\perp} = 0.3607$ nm for Cu/TiN, and $a_{||} = 0.3616$ nm and $a_{\perp} = 0.3612$ nm for Cu/MgO, which is in excellent agreement with the $a_{\perp}$ values obtained from the $\omega$-2θ scans. Using the reported relaxed Cu lattice constant of 0.3614 nm, the in-plane $\epsilon_{||}$ and orthogonal $\epsilon_{\perp}$ strains in the epitaxial Cu layers are calculated to be ($\epsilon_{||}$, $\epsilon_{\perp}$) = (0.23%, -0.19%) and (0.06%, -0.05%) for Cu/TiN with $T_s = 25$ °C and Cu/MgO with $T_s = 80$ °C, respectively, which yields Poisson ratios $\nu = (\epsilon_{\perp}/\epsilon_{||})/(\epsilon_{\perp}/\epsilon_{||} - 2) = 0.29$ and 0.30, respectively, which is close to the previously reported range of 0.31-0.36 for bulk polycrystalline copper.29,30

X-ray reflectivity is used to obtain a quantitative measurement of the Cu surface roughness over large lateral length scales. Figure 2(a) shows low-angle $\theta$-2θ reflectivity scans from epitaxial Cu layers grown on 2.5-nm-TiN/MgO(001) at $T_s = 25$ °C and on MgO(001) at $T_s = 25$ and 80 °C, with nominal Cu thicknesses of $d = 6, 7, \text{ and } 6$ nm, respectively. For this plot, the measured reflectivity $R$ is multiplied by $\theta^4$ to account for the expected angular intensity decrease from a perfectly flat surface. Thus, the rate of decrease in $R\theta^4$ versus $\theta$ is a direct measure of the surface roughness,31 while the fringes are due to interference from reflections at the Cu-air, Cu-TiN, and TiN-MgO interfaces, providing values for the average Cu layer thicknesses of $6.3 \pm 0.3, 6.8 \pm 0.3, \text{ and } 6 \pm 2$ nm for Cu/TiN/MgO with $T_s = 25$ °C, Cu/MgO $T_s = 25$ and 80 °C, respectively, which is close to the expected thicknesses of 6, 7, and 6 nm, respectively, obtained from the deposition rate and time, where the deposition rate is calibrated using Rutherford backscattering spectroscopy. The Cu/TiN/MgO sample has the smallest rate of intensity decrease and shows strong interference effects, which both indicate a relatively small surface roughness with an a root-mean-square (rms) value of 0.12 ± 0.05 nm, as determined from fitting the experimental data with the expected curves, assuming a Gaussian distribution to model surface and interface roughness and using the recursive theory of Parrat, which is based on the Fresnel reflectivity formalism.32 In contrast, the layers grown directly on

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MgO(001) show a steeper intensity decrease and less pronounced fringes. This is particularly true for $T_s = 80 \, ^\circ\text{C}$, where the absence of fringes suggests the absence of planar interfaces, indicating that this Cu layer is discontinuous. This is consistent with the value for the rms surface roughness of 4.2 nm, which is obtained from the curve fitting and corresponds to a peak-to-valley height difference of 11.8 nm, which, in turn, is 2 $\times$ larger than the nominal thickness of 6 nm. The discontinuity of this layer is also confirmed by the fact that the sheet resistance is too large to be measured by our standard 4-point probe setup, indicating a $\rho > 10 \, \Omega \cdot \text{m}$.

Figure 2(b) is a plot showing the measured Cu surface roughness versus layer thickness, $d = 6-40 \, \text{nm}$, for the three sample types of this study, including the data obtained from the scans shown in Fig. 2(a). The roughness for the Cu/TiN/MgO layers grown at $T_s = 25 \, ^\circ\text{C}$ increases from 0.12 $\pm$ 0.05 nm at $d = 6 \, \text{nm}$ to 0.8 $\pm$ 0.2 nm at $d = 40 \, \text{nm}$, which is attributed to kinetic roughening. In contrast, Cu/MgO layers with $T_s = 25 \, ^\circ\text{C}$ exhibit a roughness that remains almost constant, from 1.1 $\pm$ 0.2 nm at $d = 7 \, \text{nm}$ to 0.8 $\pm$ 0.1 nm at $d = 39 \, \text{nm}$, while the roughness of Cu/MgO grown at $T_s = 80 \, ^\circ\text{C}$ decreases from 4.2 $\pm$ 1.3 nm at $d = 6 \, \text{nm}$ to 1.2 $\pm$ 0.1 nm at $d = 40 \, \text{nm}$. The latter is attributed to the coalescence of three-dimensional nuclei to form a continuous layer with increasing $d$. In summary, the roughness in the early stages of growth ($d = 6 \, \text{nm}$) is low for growth on TiN, but is high for growth on MgO, because the lower surface energy of MgO facilitates dewetting and growth of three-dimensional nuclei, particularly when surface mass transport is enhanced at elevated $T_s$. In contrast, the thickest measured layers with $d = 39 \pm 1 \, \text{nm}$, all have an approximately equal rms surface roughness of $1 \pm 0.2 \, \text{nm}$, which is close to our earlier reported $^{33}$ rms surface roughness of 1.8 nm measured by in situ scanning tunneling microscopy for 40-nm-thick-Cu/MgO layers grown at $T_s = 80 \, ^\circ\text{C}$. $^{33}$

Figure 3 is a plot of the resistivity $\rho$ of single crystal Cu(001) layers deposited on TiN(001) and MgO(001) at 25 and 80 $^\circ\text{C}$, respectively, versus layer thickness $d = 4-1400 \, \text{nm}$, measured both in situ in vacuum at 298 K and in liquid nitrogen at 77 K. The inset shows the corresponding resistivity measured ex situ in ambient air at 298 K. The thickest Cu layers with $d = 1400 \pm 30 \, \text{nm}$ on MgO with $T_s = 80 \, ^\circ\text{C}$ and $d = 830 \pm 23 \, \text{nm}$ on TiN/MgO with $T_s = 25 \, ^\circ\text{C}$ exhibit $\rho$ values of (i) $1.70 \pm 0.06$ and $1.70 \pm 0.10 \, \mu\Omega \cdot \text{cm}$ at 298 K in vacuum, (ii) $1.71 \pm 0.10$ and $1.71 \pm 0.10 \, \mu\Omega \cdot \text{cm}$ at 298 K in air, and (iii) $0.20 \pm 0.01$ and $0.21 \pm 0.01 \, \mu\Omega \cdot \text{cm}$ at 77 K, respectively. These values are close to the expected Cu bulk resistivity at the respective temperatures, $\rho_{\text{Cu}}^{77K} = 1.712 \, \mu\Omega \cdot \text{cm}$ and $\rho_{\text{Cu}}^{77K} = 0.213 \, \mu\Omega \cdot \text{cm}$. As $d$ decreases to 20.0 $\pm$ 0.4 nm, the resistivity of the Cu/MgO layers with $T_s = 80 ^\circ\text{C}$ increases to reach $2.65 \pm 0.09$, $2.94 \pm 0.09$, and $0.96 \pm 0.03 \, \mu\Omega \cdot \text{cm}$ in vacuum, air, and liquid nitrogen, respectively. The resistivity of the corresponding layers with nominal thicknesses, $d = 10$ and 7 nm (not shown), is too high to be measured, which is attributed to these Cu layers being discontinuous, which is consistent with the x-ray reflectivity results. On the contrary, the resistivity of the Cu/TiN/MgO layers could be measured for all thicknesses. For example, the layer with $d = 6.2 \pm 0.2$ nm exhibits $\rho = 4.4 \pm 0.4$, $6.9 \pm 0.4$, and 2.8 $\pm$ 0.2 $\mu\Omega \cdot \text{cm}$ in vacuum, air, and liquid nitrogen, respectively. The resistivity increase with decreasing $d$ is attributed to electron scattering at the bottom and top surfaces, where the bottom surface corresponds to Cu/MgO and Cu-TiN interfaces and the top surface corresponds to the Cu-vacuum, Cu-air, and Cu-liquid N$_2$ boundary for measurements at 298 (vacuum), 298 (air), and 77 K, respectively. The difference in the resistivity for a given sample measured at 298 K (vacuum) and 77 K is independent of $d$ and is $1.50 \pm 0.10 \, \mu\Omega \cdot \text{cm}$, which is in exact agreement with $\rho_{\text{Cu}}^{298K} - \rho_{\text{Cu}}^{77K} = 1.499 \, \mu\Omega \cdot \text{cm}$, suggesting that the effect of electron scattering at surfaces and phonons is additive, in agreement with Matthiessen’s rule. The resistivity of Cu/MgO layers with $T_s = 25 \, ^\circ\text{C}$ (not shown) at $d = 40$ and 6 nm is $2.6 \pm 0.2$ and $6.4 \pm 0.5 \, \mu\Omega \cdot \text{cm}$ at 298 K in vacuum, respectively, which is $\sim 1.4 \times$ larger than the respective Cu/TiN layers with $T_s = 25 \, ^\circ\text{C}$ and is attributed to additional electron scattering by crystalline defects, consistent with the lowest crystalline quality observed by XRD for the Cu/MgO layers grown at $T_s = 25 \, ^\circ\text{C}$.

The resistivity measurements of all Cu layers are discussed within the theoretical framework by Fuchs $^{34}$ and Sondheimer $^{35}$ (FS) using the equation we reported earlier $^{36}$ that integrates over all classical electron paths to predict the effect of surface scattering on the resistivity $\rho$ for a thin film of thickness $d$ with different specularity parameters $p_1$ and $p_2$ for the top and bottom surfaces, respectively.

$$\rho_s = \rho_0 \left[ 1 - \frac{3}{4\pi} \int_1^\infty \left( \frac{1}{r^2} - \frac{1}{r_s^2} \right) \frac{2 \left( 1 - p_1 p_2 e^{-r_s} \right) \left( 1 - e^{-r_s} \right) - \left( p_1 + p_2 \right) \left( 1 - e^{-r_s} \right)^2 dt \right]^{-1}, \right. \quad \text{(1)}$$
where \( \kappa = d/\lambda \). Here, \( \lambda \) is calculated using the Fermi free electron model and the experimental bulk resistivity \( \rho_0 \). The lines in Fig. 3 show the expected resistivity obtained by numerically integrating Eq. (1), using \( \lambda = 39 \) and 313 nm for 298 and 77 K, respectively, and using a specularity parameter for the top-surface of \( p_1 = 0, 0.5, \) and 1, as labeled, and for the bottom surface of \( p_2 = 0 \). The \( p_2 \) is set to zero, because the Cu-MgO and Cu-TiN interfaces are expected to yield completely diffuse electron scattering, since the \( \rho \) for air-exposed Cu/MgO and Cu/TiN layers follow the prediction from Eq. (1) for completely diffuse scattering on both surfaces \( (p_1 = p_2 = 0 \pm 0.1) \), as shown in the inset of Fig. 3, which is also consistent with Refs. 26 and 36. All measured data points at 298 K (vacuum) and 77 K are below the \( p_1 = 0 \) line, indicating that the top surface exhibits partial specular scattering with an average, \( p_1 = 0.6 \pm 0.2 \), which agrees with our previously reported \( p_1 = 0.6 \pm 0.1 \) for Cu/MgO layers.36 The slight differences in \( p_1 \) values at 298 K (vacuum) and 77 K for a specific sample is within the experimental uncertainty but may also indicate enhanced or reduced specular scattering due to effective flattening or roughening of the electrostatic surface potential associated with polarized physisorbed nitrogen on the Cu surface.37 We also note that the FS model in Eq. (1) deviates from Mathieson’s rule for small \( d \). However, the difference is comparable in magnitude to the experimental uncertainty, so that this deviation cannot be observed with any certainty from the experiment.

IV. CONCLUSIONS

X-ray diffraction and reflection analyses show that Cu(001) layers grown on MgO(001) are single crystals, but exhibit a relatively low crystalline quality with a Cu 002 rocking curve width of 4.7 and 3.8° for \( d = 40 \) nm and \( T_s = 25 \) and 80 °C, respectively, along with a relatively high surface roughness at a small thickness \( d = 6-7 \) nm of 1.1 and 4.2 nm, respectively. Thus, increasing \( T_s \) from 25 to 80 °C results in a 1.2 × higher crystalline quality, but a 4 × larger surface roughness for thin Cu(001)/MgO(001) layers. However, the deposition of a 2.5-nm-thick TiN(001) layer prior to Cu deposition yields a 3.5 × narrower Cu 002 rocking curve width than when grown directly on MgO at \( T_s = 80 \) °C, and a 9 × smaller surface roughness than Cu/MgO layers with \( T_s = 25 \) °C, for the same Cu layer thicknesses. This suggests that TiN effectively acts as a wetting layer on MgO(001) and provides an approach for the growth of continuous single crystal Cu(001) layers with thicknesses as low as 4 nm, which are well suited to studying electron surface scattering. The layers were found to exhibit a FS specularity parameter at the Cu-vacuum interface of 0.6 ± 0.2, however, they exhibit a value of 0 ± 0.1 for the Cu-air interface.

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