

Effects of Thermal N₂ Annealing on Passivation Capability of Sputtered Ta(-N) Layers Against Cu Oxidation

Jui-Chang Chuang and Mao-Chieh Chen*

Department of Electronics Engineering and Institute of Electronics, National Chiao-Tung University,
Hsinchu 300, Taiwan

ABSTRACT

The effects of thermal N₂ annealing on the passivation capability of sputtered Ta and Ta-nitride [Ta(-N)] layers against Cu oxidation in a 200 Å Ta(-N) covered Ta(-N)/Cu/SiO₂/Si structures was investigated. The N₂ annealed Ta layers revealed degradation in passivation capability, presumably due to grain growth of the Ta passivation layer. In contrast, the nitrogen-doped Ta-nitride layers showed a contrary trend. For the Ta-nitride layer with 23.5 atom % of nitrogen, passivation capability was effectively improved by N₂ annealing at 300°C. For the Ta-nitride layer with 30.5 atom % of nitrogen, N₂ annealing at higher temperatures (500–700°C) was necessary to improve the passivation capability. The healing of sputtering damage of Ta-nitride passivation layers by the thermal N₂ annealing was responsible, presumably, for the improvement of passivation capability.

Introduction

Copper has been extensively studied as a potential substitute for aluminum and Al alloys in multilevel metallization of semiconductor devices.^{1–3} Compared with Al and Al alloys, Cu has a number of beneficial factors, such as lower bulk resistivity (1.7 vs. 2.7 and >4.0 mΩ cm),^{1,4} higher electromigration resistance,⁵ higher melting point,^{4,6} and lower reactivity with commonly used diffusion barrier materials.^{6,7} However, there are obstructive factors compared with Al and Al alloys, such as difficulty in dry etching,⁸ poor adhesion to the dielectric layer,^{1,2} easy diffusion in silicon and SiO₂,^{9–13} deep level trap in silicon, and Cu silicide formation at low temperatures. Thus, the use of barrier layers to reduce the diffusion of Cu and to improve its adhesion to dielectrics, especially SiO₂,^{8,14–20} is of great importance. The various barrier/adhesion layers for Cu metallization that have been extensively investigated are mostly listed in Ref. 14.

It is well known that Cu oxidizes easily in air and in humid ambient,^{21–24} even at room temperature. This characteristic has impeded the application of Cu in integrated circuits. A proper technique of passivation against Cu oxidation must be developed for its widespread application.²⁵ So far, passivation schemes used to resist Cu oxidation in an oxidizing ambient have been studied^{1,2} and proposed by some researchers, among them the self-aligned passivation by Al, Mg,^{26,27} Ti, Cr,²⁸ and Nb,²⁹ the sidewall passivation^{30,31} by Mo and TiN, the formation of surface silicide,³² and the B-implanted Cu.³³

Previously, the passivation capability of 200 Å tantalum (Ta) and tantalum nitride (Ta₂N or Ta₃N) [or for short Ta(-N)] covered Cu films were studied and found by the authors to be able to withstand thermal annealing in flowing O₂ at temperatures up to 400°C without Cu oxidation.³⁴ In the current study, electrical measurement and various techniques of material analysis are used to study the effects of thermal N₂ annealing of Ta(-N)/Cu/SiO₂/Si structures on the passivation capability of Ta(-N) layers against Cu oxidation.

Experimental

For sample preparation, the starting materials were p-type, boron-doped, 3 in. diam Si wafers with nominal resistivity of 17–55 Ω cm. After initial RCA cleaning, the Si wafers were thermally oxidized at 1050°C in steam to grow a 5000 Å SiO₂. A Cu film 2000 Å thick was sputter deposited on the oxide layer, which was followed by a 200 Å Ta or Ta-nitride film deposition on the Cu film in the same sputtering system without breaking the vacuum. The Ta film was sputter deposited using a pure Ta target (99.999% purity) in Ar ambient, while the Ta-nitride films were deposited by reactive sputtering using the same Ta target in a gas mixture of Ar and N₂ with various flow rates. The

gases used were of electronic grade, and the base pressure of the deposition chamber was 5×10^{-7} Torr. All films were sputtered at a pressure of 7.8 mTorr without intentional substrate heating. Table I summarizes the sputtering condition and sample identification; the percentage atomic concentrations of nitrogen in the sputtered films are also tabulated.³⁵

Then, the wafers were diced into 1.5×1.5 cm pieces for various thermal treatments. First, the diced wafers were thermally treated in N₂ ambient for 30 min at 300, 500, and 700°C separately; this thermal process is termed thermal N₂ annealing. To study the effect of N₂ annealing on the passivation capability, the as-deposited as well as the N₂ annealed samples were thermally annealed in flowing O₂ ambient for 50 min at various temperatures. The sample identifications are summarized in Table II. A_{as}350O, for example, is designated as sample A, thermally annealed in O₂ ambient at 350°C without any N₂ annealing, and B_{50N}450O is designated as sample B, annealed at 500°C in N₂ ambient followed by thermal annealing in O₂ ambient at 450°C.

Electrical measurement and material analysis were used to characterize the passivation capability of as-deposited and N₂ annealed Ta and Ta-nitride films. Sheet resistance (R_s) was measured using a four-point probe. Scanning electron microscopy (SEM) was used to investigate surface morphology, X-ray photoelectron spectroscopy (XPS) was used for chemical state analysis, X-ray diffraction (XRD) analysis was used for phase identification, and secondary ion mass spectroscopy (SIMS) and Auger electron spectroscopy (AES) were used for elemental depth profile analysis.

Results and Discussion

The passivation capability of 200 Å Ta and Ta-nitride layers for the Ta(-N) covered Ta(-N)/Cu/SiO₂/Si structures in an oxidizing ambient was investigated in a previous study by the authors.³⁴ It was also found that N₂ annealing of these Ta(-N) layers did not change the chemical states and the atomic concentrations of tantalum and nitrogen.³⁵ In the current study, the effects of N₂ annealing on the passivation capability of the Ta(-N) layers were examined.

Table I. Sputtering condition and nitrogen content of as-deposited Ta and Ta-nitride films.

Sample identification	A (Ta)	B (Ta-nitride)	C (Ta-nitride)
Partial pressure of N ₂ in sputtering gas (mTorr)	0	1.52	2.4
Deposition power (W)	150	150	150
Deposition rate (Å/s)	0.3	0.3	0.3
Nitrogen atomic concentration (%)	0	23.5	30.5

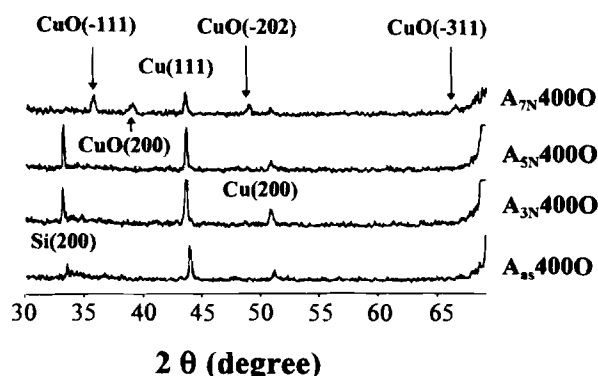
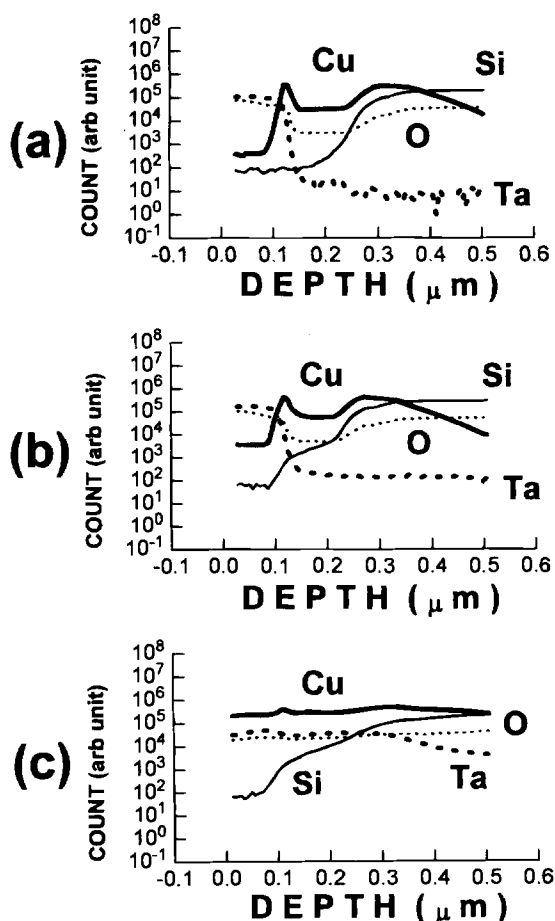
* Electrochemical Society Active Member.

Table II. Summary of sample identification and N₂/O₂ annealing temperatures.

A samples					B samples					C samples				
N ₂ annealing	as-deposited	A _{as}				B _{as}				C _{as}				
	300°C	A _{3N}				B _{3N}				C _{3N}				
	500°C	A _{5N}				B _{5N}				C _{5N}				
	700°C	A _{7N}				B _{7N}				C _{7N}				
O ₂ annealing	350°C	A _{as} 350O	A _{3N} 350O	A _{5N} 350O	A _{7N} 350O	B _{as} 350O	B _{3N} 350O	B _{5N} 350O	B _{7N} 350O	C _{as} 350O	C _{3N} 350O	C _{5N} 350O	C _{7N} 350O	
	400°C	A _{as} 400O	A _{3N} 400O	A _{5N} 400O	A _{7N} 400O	B _{as} 400O	B _{3N} 400O	B _{5N} 400O	B _{7N} 400O	C _{as} 400O	C _{3N} 400O	C _{5N} 400O	C _{7N} 400O	
	450°C	A _{as} 450O	A _{3N} 450O	A _{5N} 450O	A _{7N} 450O	B _{as} 450O	B _{3N} 450O	B _{5N} 450O	B _{7N} 450O	C _{as} 450O	C _{3N} 450O	C _{5N} 450O	C _{7N} 450O	
	500°C	A _{as} 500O	A _{3N} 500O	A _{5N} 500O	A _{7N} 500O	B _{as} 500O	B _{3N} 500O	B _{5N} 500O	B _{7N} 500O	C _{as} 500O	C _{3N} 500O	C _{5N} 500O	C _{7N} 500O	

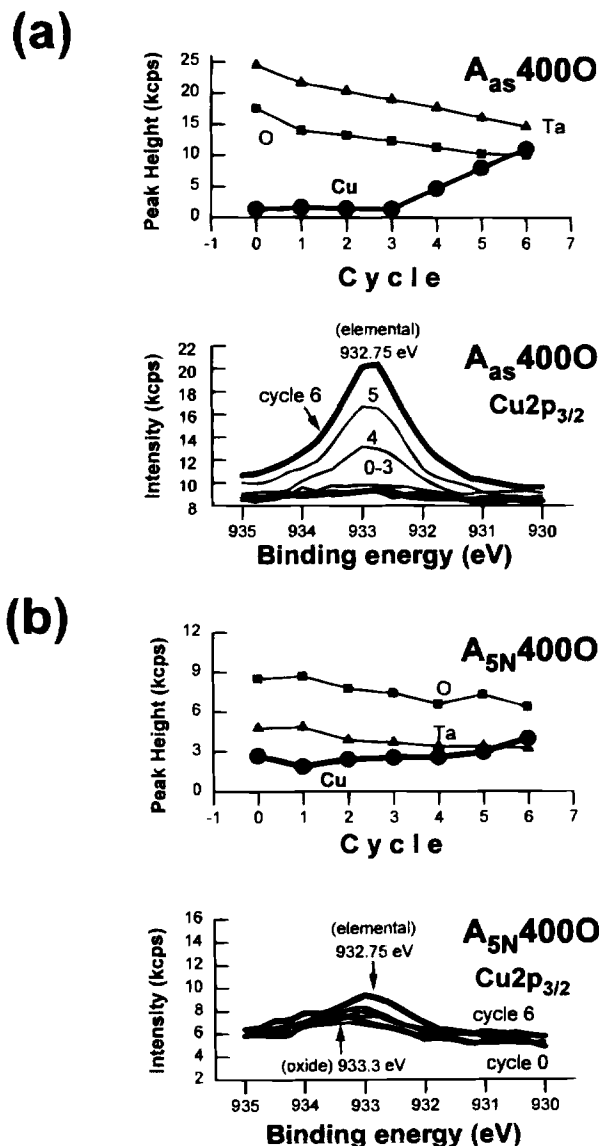
Ta Passivation Layers

A samples.—Figure 1 shows the XRD spectra of 400°C O₂ annealed A samples. Signals of the CuO phase appeared only in the spectrum of A_{7N}400O sample. Also, the sheet

Fig. 1. XRD spectra of 400°C O₂ annealed A samples.Fig. 2. SIMS depth profiles of compositional elements for the samples (a) A_{as}400O, (b) A_{5N}400O, and (c) A_{7N}400O.

resistance measured of this sample was eight times larger than that of the other samples, indicating the oxidation³⁴ of Cu in this sample.

Figure 2 shows the SIMS depth profiles of A samples. There was not much structural difference between samples A_{as}400O (Fig. 2a) and A_{5N}400O (Fig. 2b). However, the A_{7N}400O samples showed a significant change in elemental depth profiles (Fig. 2c), indicating the degradation of passivation capability resulting from a 700°C N₂ annealing. Figure 3 shows the XPS depth profiles of compositional elements and the binding energy spectra of Cu 2p_{3/2} photoelectrons for the A_{as}400O and A_{5N}400O samples. A

Fig. 3. XPS depth profiles of compositional elements and XPS binding energy spectra of Cu 2p_{3/2} photoelectrons for the samples (a) A_{as}400O and (b) A_{5N}400O.

copper oxide phase of CuO was identified in the spectrum of $A_{5N}400O$ (Fig. 3b, cycles 0) but not in the spectrum of $A_{as}400O$ (Fig. 3a). The degradation of passivation capability for the Ta passivation layer due to N_2 annealing can be more clearly observed on the SEM micrographs shown in Fig. 4, in which CuO clusters^{21-24,36} are present on the surface of N_2 annealed samples. The copper oxide phase of CuO was also determined by the XPS analysis for the sample of $A_{7N}400O$. In fact, the degradation occurred due to N_2 annealing at temperatures as low as 300°C.

The degradation of passivation capability of Ta passivation layer is presumably caused by the grain growth of the thin Ta passivation layer. Figures 5 and 6 show, respectively, the XRD spectra and the surface morphology for A

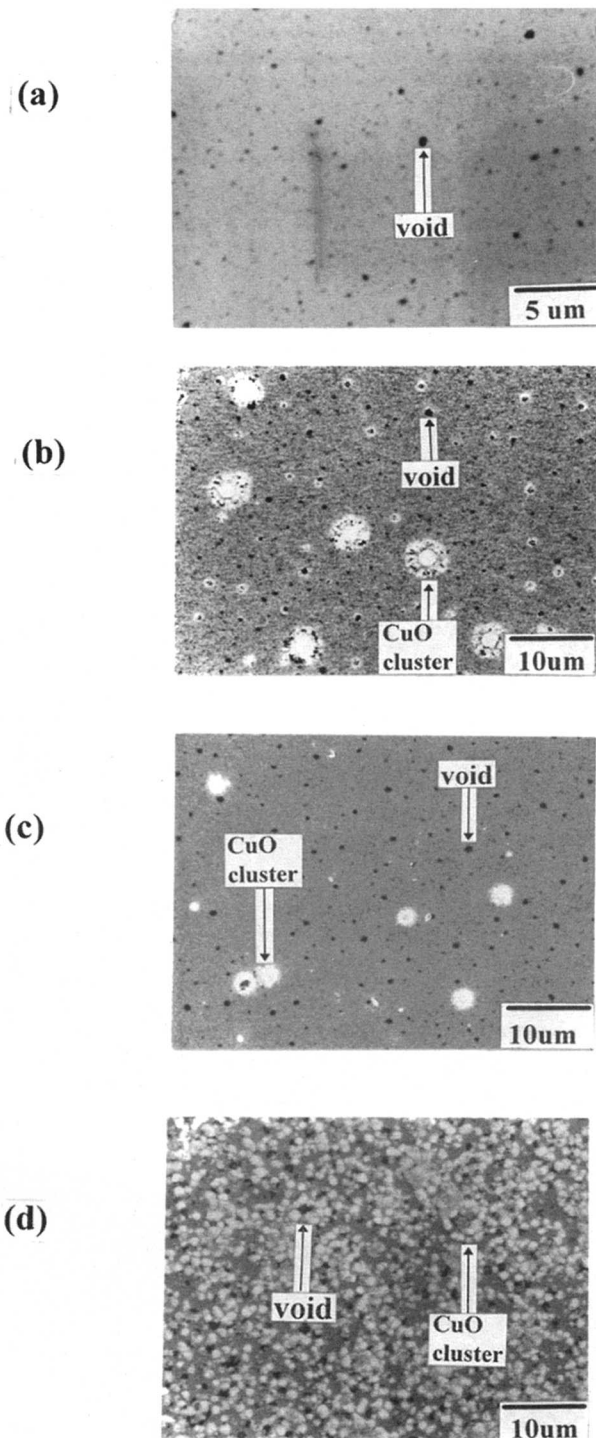


Fig. 4. SEM micrographs showing surface morphology for the samples (a) $A_{as}400O$, (b) $A_{3N}400O$, (c) $A_{5N}400O$, and (d) $A_{7N}400O$.

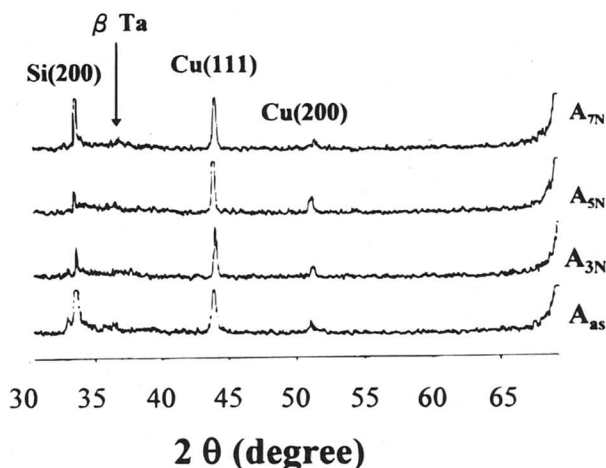


Fig. 5. XRD spectra for A samples with and without N_2 annealing.

samples with and without N_2 annealing. The βTa signal was detected on sample A_{7N} (Fig. 5) and Ta flaking was observed on the surface of sample A_{5N} as well as A_{7N} (Fig. 6b). As the grain size became compatible with the film thickness, shorter diffusion paths^{14-20,34-37} were built along the grain boundaries and the voids,^{34,35} leading to degradation of passivation capability.

Ta-Nitride Passivation Layers

B Samples.—Figure 7 shows the XRD spectra of 450°C O_2 annealed B samples. No copper oxide phase was observed by the XRD analysis. After the O_2 annealing at 450°C, the sheet resistance remained constant for all B samples except sample $B_{as}450O$, which showed a 150% increase in sheet resistance as compared with that of sample B_{as} . Figure 8 illustrates the AES depth profiles of as-deposited and 450°C O_2 annealed B samples. The passivated Cu film and the Ta-nitride passivation layer for the N_2 annealed samples Fig. 8c, d, and e remained basically unchanged as

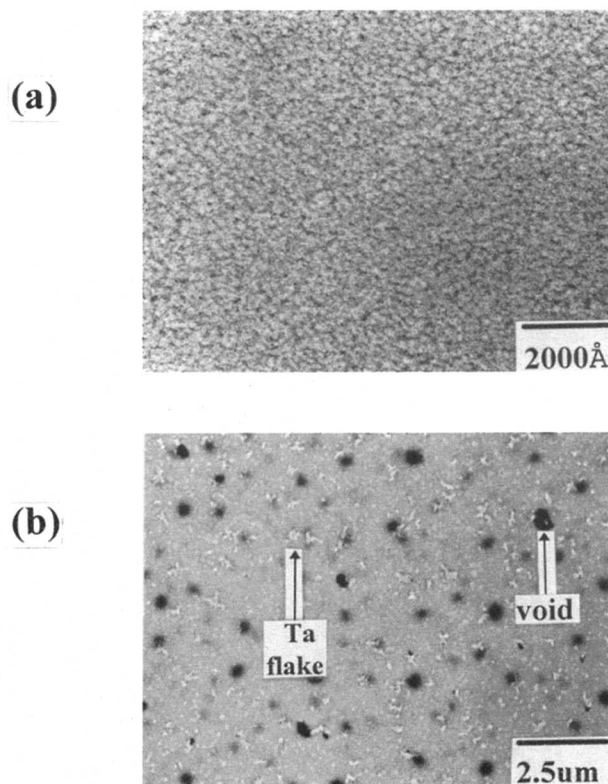


Fig. 6. SEM micrographs showing surface morphology for the samples (a) A_{as} and (b) A_{7N} .

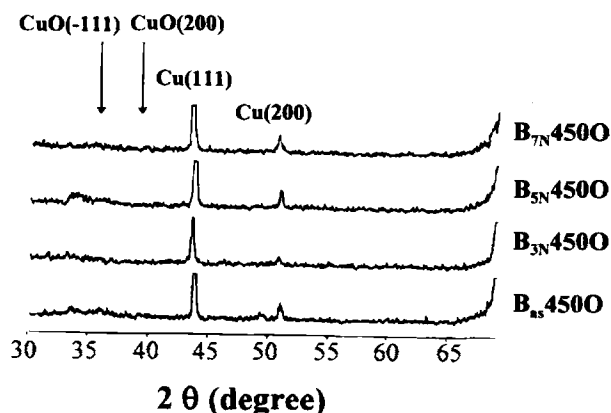


Fig. 7. XRD spectra of 450°C O₂ annealed B samples.

compared with those of the as-deposited sample B_{as} (Fig. 8a); however, complete intermixing of Cu, Ta, and O occurred for the sample of B_{as}450O (Fig. 8b), which did not go through N₂ annealing. Figure 9 shows the results of the XPS analysis. The Cu 2p_{3/2} photoelectrons were detected on the surface of sample B_{as}450O (Fig. 9a), and were identified to be in the state of CuO. The XPS spectra of Cu 2p_{3/2} photoelectrons for sample B_{7N}450O in the surface region and under the Ta-nitride layer are illustrated in Fig. 9b. No detectable Cu photoelectron was found in the surface region and the Ta-nitride layer. The Cu 2p_{3/2} photoelectrons found under the Ta-nitride passivation layer were determined to be in the elemental state. For sample B_{7N}450O, the only oxide phase detected by the XPS was Ta₂O₅ existing in the outermost surface region. Under the Ta₂O₅ layer, Ta 4f_{7/2} and N 1s photoelectrons remained in their nitride phase as were in the as-deposited sample.³⁵

The N₂ annealing process for B samples effectively improved the passivation capability of their passivation lay-

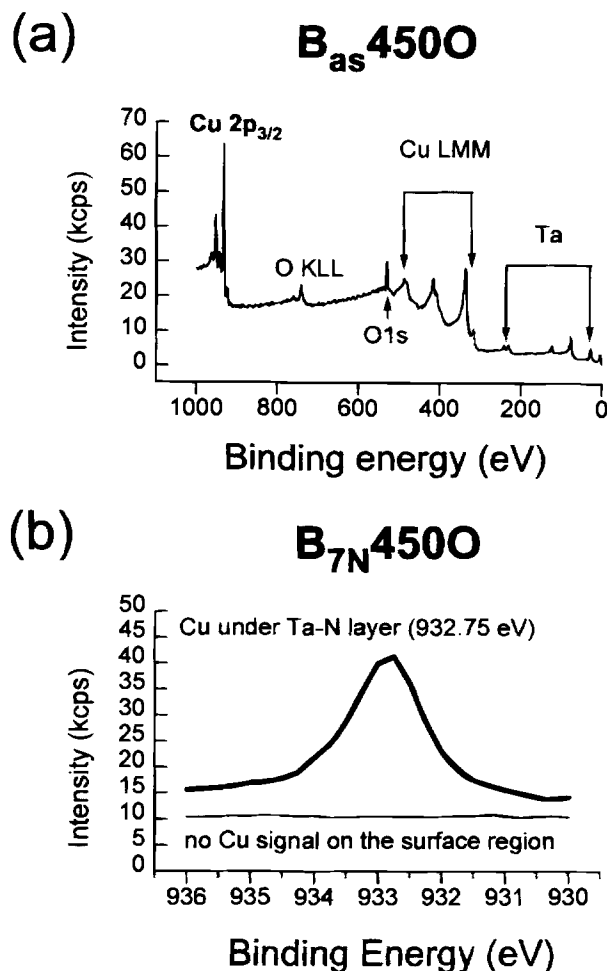


Fig. 9. XPS spectra showing (a) surface compositional elements survey for the sample B_{as}450O and (b) binding energy spectrum of Cu 2p_{3/2} photoelectrons for the sample B_{7N}450O on the outermost surface as well as under the Ta-nitride passivation layer.

ers. And the superiority of passivation capability of sample B_{as} over that of sample A_{as} was presumably due to amorphism^{34-36,38} of the Ta-nitride layer. Figure 10 shows the surface morphology for the as-deposited Ta-nitride (or Ta₂N)³⁶ layer (Fig. 10a) as well as the 700°C N₂ annealed film (Fig. 10b). No obvious grain growth of Ta-nitride was observed. This implies the fine grain of the Ta-nitride film and thus a long diffusion path for Cu along the grain boundaries. The thermal N₂ annealing healed the micro-damage in the reactively sputtered Ta-nitride layer;³⁵ however, it resulted in new defects, i.e., void formation, as shown in Fig. 6b and 10b. Thus, the passivation improvement was limited to thermal annealing in O₂ ambient at 450°C. It was found that none of samples A or B, either with or without N₂ annealing, can withstand a thermal annealing in O₂ ambient at 500°C for 50 min without the Cu film being oxidized.

C Samples.—Figure 11 shows the XRD spectra of 400°C O₂ annealed C samples. Signals of the CuO phase were detected on sample C_{as}400O, which did not go through any N₂ annealing process. For the sample annealed in N₂ ambient at 500°C, the signals of CuO diminished. For the sample annealed in N₂ ambient at 700 and 800°C, a CuO phase signal was not observed in the XRD spectrum. Figure 12 shows the percentage change of the sheet resistance resulting from thermal O₂ annealing vs. temperatures of N₂ annealing for C samples thermally annealed in O₂ ambient at 350 and 400°C. The improvement of passivation capability resulting from N₂ annealing was indicated by the abrupt decrease of the sheet resistance change. For thermal O₂ annealing at 350°C, N₂ annealing at temperatures above 500°C was needed; for

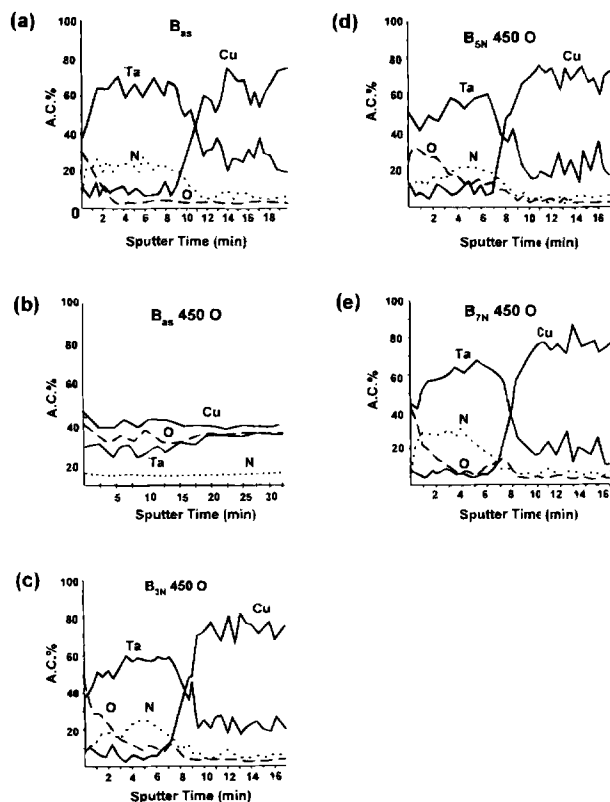


Fig. 8. AES depth profiles of compositional elements for the samples (a) B_{as}, (b) B_{as}450O, (c) B_{3N}450O, (d) B_{5N}450O, and (e) B_{7N}450O.

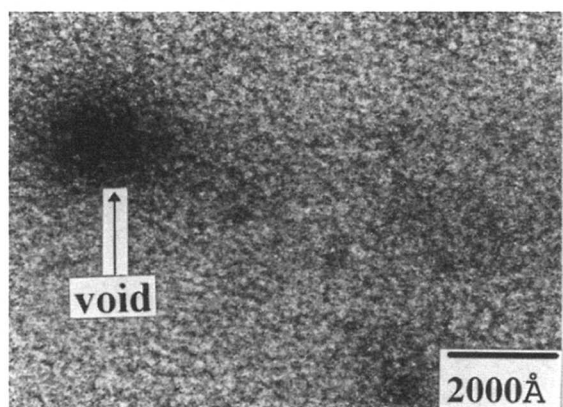
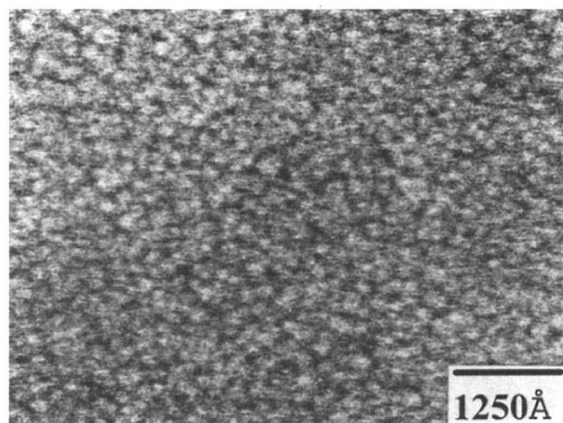


Fig. 10. SEM micrographs showing surface morphology for the samples (a, top) B_{as} and (b, bottom) B_{7N} .

the thermal O_2 annealing at $400^\circ C$, N_2 annealing at temperatures above $700^\circ C$ was needed. Figure 13 illustrates the AES depth profiles for C samples thermally annealed in O_2 ambient at $350^\circ C$. For samples $C_{as}350O$ and $C_{5N}350O$, Cu was oxidized partially and the Ta-nitride passivation layer was covered with a layer of CuO (Fig. 13a and b). The extent of Cu oxidation was reduced by raising the N_2 annealing temperature to $700^\circ C$ (Fig. 13c). Figure 14 shows the SIMS depth profiles for C samples thermally annealed in O_2 ambient at $400^\circ C$. For sample $C_{as}400O$, the original structure of Ta-nitride/Cu/ SiO_2 /Si was destroyed and copper oxide can be seen clearly (Fig. 14a). Similar results were observed for the samples with N_2 annealing at temperatures below $700^\circ C$. For samples $C_{7N}400O$ and $C_{8N}400O$, which were annealed in N_2 at 700 and $800^\circ C$, respectively, the original structure of Ta-nitride/Cu/ SiO_2 /Si retained its integrity (Fig. 14b and c). Figure 15 shows the XPS survey spectra of surface elements for the C samples. Only the samples annealed in N_2 at $700^\circ C$ did not reveal Cu 2p photoelectrons which were determined to be CuO phase. From the above

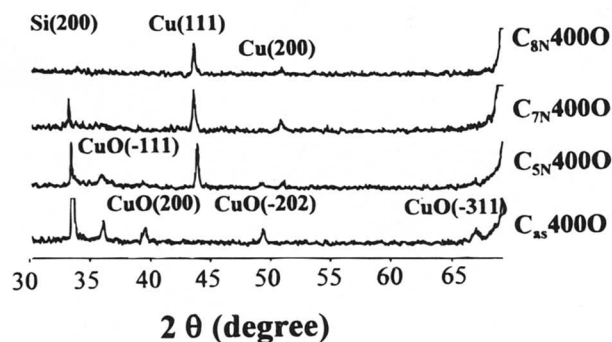


Fig. 11. XRD spectra of $400^\circ C$ annealed C samples.

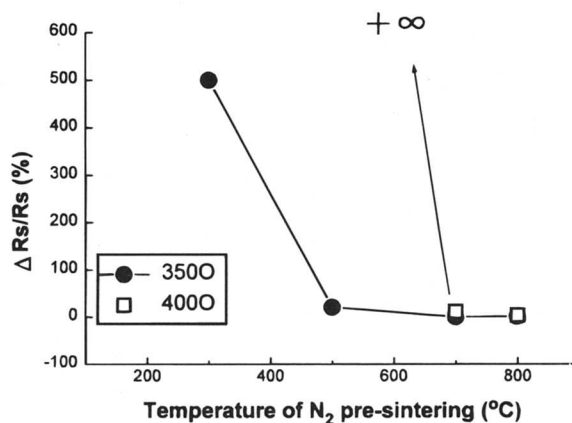


Fig. 12. Sheet resistance change ($\Delta R_s/R_s$) vs. temperature of N_2 annealing for C samples thermally annealed in O_2 ambient at $350^\circ C$ and $400^\circ C$.

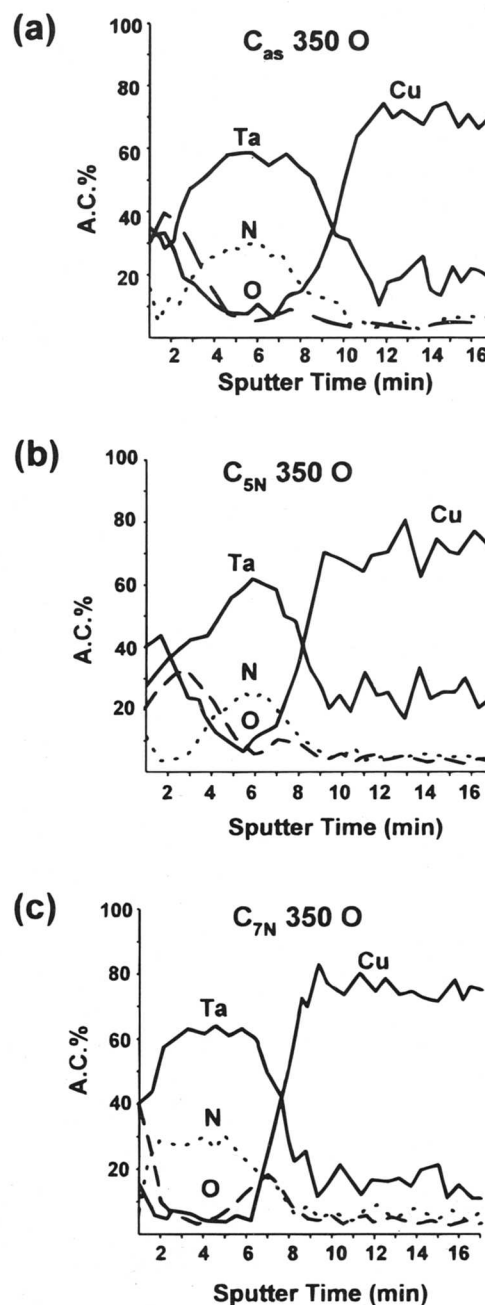


Fig. 13. AES depth profiles of compositional elements for the samples (a) $C_{as}350O$, (b) $C_{5N}350O$, and (c) $C_{7N}350O$.

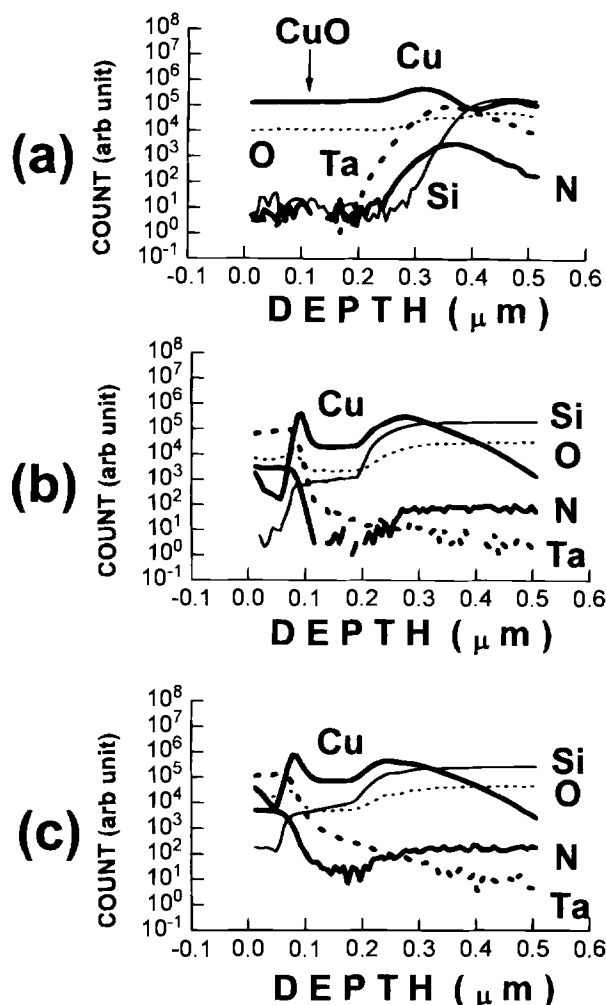


Fig. 14. SIMS depth profiles of compositional elements for the samples (a) $C_{ox}400O$, (b) $C_{7N}400O$, and (c) $C_{8N}400O$.

observation, we conclude that the passivation capability of C samples, can be improved by N_2 annealing at temperatures above $700^\circ C$.

Figure 16 shows the XRD spectra of as-deposited and thermally N_2 annealed C samples. A small peak belonging to the Ta-nitride phase appeared in the C_{7N} spectrum. Similar to B samples, the thermal N_2 annealing did not result in the change of atomic concentration and chemical state of Ta and N in C samples.³⁵ A comparison of the surface morphologies of the as-deposited sample C_{as} and the $700^\circ C$ N_2 annealed sample C_{7N} is shown in Fig. 17. N_2 annealing at $700^\circ C$ resulted in no obvious grain growth. The amorphousness of the Ta-nitride films of C samples is similar to that of B samples. However, the void formation (Fig. 17b) prohibited further improvement on the passivation capability by thermal N_2 annealing at higher temperatures.

To improve passivation capability, an N_2 annealing at $300^\circ C$ would be beneficial to sample B, but a $700^\circ C$ N_2 annealing was necessary for sample C to be effective. The difference lies in the degree of sputtering damage between sample B_{as} and sample C_{as} . For the as-deposited sample C_{as} , visible damage resulting from energetic N_2 plasma during sputtering was observed (Fig. 17a). We presumed that the sputtering induced damage^{34,35} was responsible for the inferior passivation capability of sample C_{as} as compared with sample B_{as} . The improvement of passivation capability for the high temperature N_2 annealed C samples was principally due to sputtering damage healing.

Conclusion

This work studied the effects of thermal N_2 annealing on the passivation capability of sputtered Ta and Ta-nitride

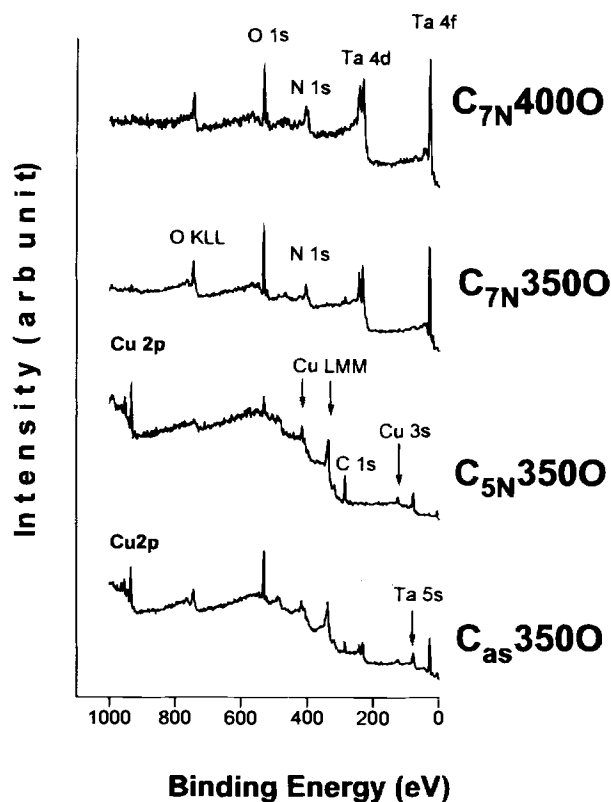


Fig. 15. XPS survey spectra of surface compositional elements for samples $C_{ox}350O$, $C_{5N}350O$, $C_{7N}350O$, and $C_{7N}400O$.

layers against Cu oxidation in a $200 \text{ \AA}$ Ta(-N) covered Ta(-N)/Cu/SiO₂/Si structure. For Ta passivation layers, the Ta/Cu/SiO₂/Si structure was able to sustain a thermal annealing in O_2 ambient at $400^\circ C$ for 50 min without Cu oxidation. However, N_2 annealing at temperatures of $300^\circ C$ and above resulted in the degradation of passivation capability. This is, presumably, because thermal N_2 annealing resulted in grain growth for the Ta passivation layer, leading to shorter paths of diffusion for Cu along the grain boundaries. The N-doped Ta-nitride layers revealed an opposite effect on passivation capability with respect to N_2 annealing. For the Ta-nitride layer of sample B (with 23.5 atom % nitrogen in Ta-nitride), passivation capability was effectively improved by the N_2 annealing process at $300^\circ C$. The N_2 annealed sample was able to sustain thermal annealing in O_2 ambient at temperatures up to $450^\circ C$ without Cu oxidation. Moreover, the higher the N_2 annealing temperature was, the better the result of passivation became. For the Ta-nitride layer of sample C (with 30.5

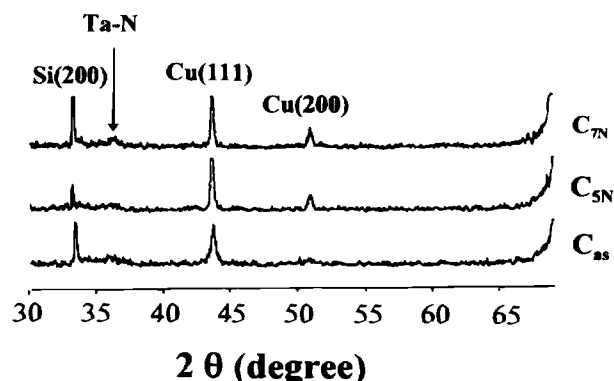


Fig. 16. XRD spectra of as-deposited and thermally N_2 annealed C samples.

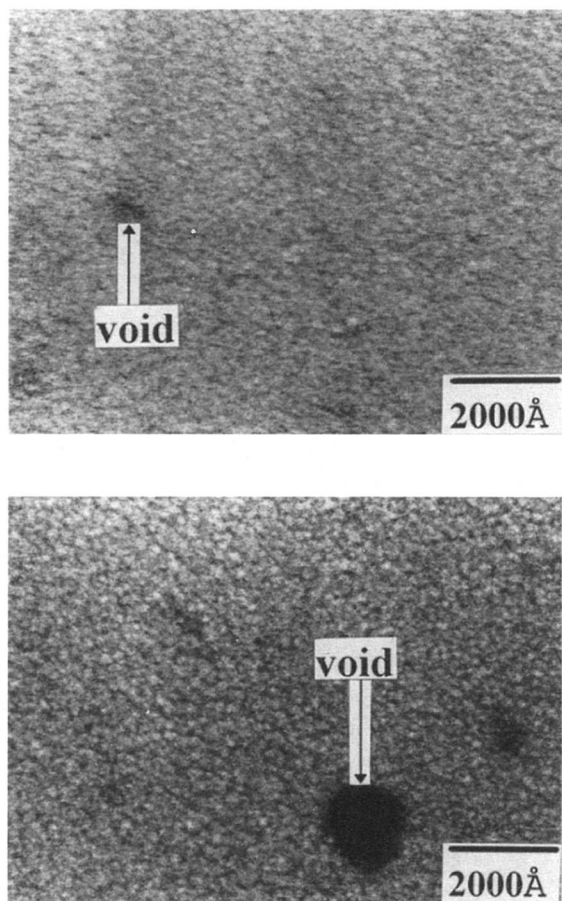


Fig. 17. SEM micrographs showing surface morphology for the samples (a, top) C_{65} and (b, bottom) C_{7N} .

atom % nitrogen in Ta-nitride), N_2 annealing at higher temperatures of 500 to 700°C was needed for the improvement to be effective because the as-deposited Ta-nitride layer was more seriously damaged. The sputtering damage healing of Ta-nitride passivation layers of sample B and sample C by the thermal N_2 annealing was, presumably, responsible for the improvement of passivation capability.

Acknowledgment

The authors wish to thank the Semiconductor Research Center of National Chiao-Tung University for providing an excellent processing environment. This work was supported by the National Science Council, ROC, under contract NSC-86-2215-E-009-040.

Manuscript submitted October 6, 1997; revised manuscript received August 14, 1998.

National Chiao-Tung University assisted in meeting the publication costs of this article.

REFERENCES

- See, for example, *MRS Bull.*, **18** (1993) and **19** (1994), devoted to the topic of copper metallization for microelectronics.
- See, for example, *Thin Solid Films*, **262** (1995), devoted to copper-based metallization and interconnects for ultralarge-scale integration applications.
- K. P. Rodbell, E. G. Colgan, and C.-K. Hu, *Mater. Res. Soc. Symp. Proc.*, **337**, 59 (1994).
- CRC Handbook of Chemistry and Physics*, 73rd ed., D. R. Lide, Editor, Section 12, CRC Press, Inc., Boca Raton, FL (1992).
- S. Shingubara, K. Fujiki, A. Sano, H. Sakaue, and Y. Horiike, *Mater. Res. Soc. Symp. Proc.*, **338**, p. 441, P. Børgesen, J. C. Coburn, J. E. Sánchez, Jr., K. P. Rodbell, and W. F. Filter, Editors, Materials Research Society, Pittsburgh, PA (1994).
- Binary Alloy Phase Diagrams*, 2nd ed. T. B. Massalski, Editor, pp. 1446, 1485, 2703, and 2919, ASM International, Material Park, OH (1990).
- E. A. Brandes, *Smithells Metals Handbook*, 6th ed., Chap. 8, p. 21, Robert Hartnoll Ltd., Bodmin, Cornwall, England (1983).
- Y. Igarashi, T. Yamanobe, and T. Ito, *Thin Solid Films*, **262**, 124 (1995).
- J. D. McBrayer, R. M. Swanson, and T. W. Sigmon, *J. Electrochem. Soc.*, **133**, 1242 (1986).
- Y. Shacham-Diamand, A. Dedhia, D. Hoffstetter, and W. G. Oldham, *J. Electrochem. Soc.*, **140**, 2427 (1993).
- D. Gupta, *Mater. Res. Soc. Symp. Proc.*, **337**, p. 209 (1994), S. P. Murarka, A. Katz, K. N. Tu, and K. Maex, Editors, p. 209, Materials Research Society, Pittsburgh, PA (1994).
- G. Raghaven, C. Chiang, P. B. Anders, S. M. Tzeng, R. Villasol, G. Bai, M. Bohr, and D. B. Fraser, *Thin Solid Films*, **262**, 168 (1995).
- S. M. Sze, *Physics of Semiconductor Devices*, 2nd ed., Chap. 6-8, John Wiley & Sons, New York (1981).
- S. Q. Wang, S. Suthar, C. Hoefflich, and B. J. Burrow, *J. Appl. Phys.*, **73**, 2301 (1993).
- J.-C. Chiou, K.-C. Juang, and M.-C. Chen, *J. Electrochem. Soc.*, **142**, 2326 (1995).
- J. S. Reid, E. Kolawa, R. P. Ruiz, and M.-A. Nicolet, *Thin Solid Films*, **236**, 319 (1993).
- H. Ono, T. Nakano, and T. Ohta, *Appl. Phys. Lett.*, **64**, 1511 (1994).
- K.-C. Park and K.-B. Kim, *J. Electrochem. Soc.*, **142**, 3110 (1995).
- J. O. Olowolafe, J. Li, J. W. Mayer, and E. G. Colgan, *Appl. Phys. Lett.*, **58**, 469 (1991).
- A. Noya, M. Takeyama, K. Sasaki, E. Aoyagi, and K. Hiraga, *Jpn. J. Appl. Phys.*, Part 1, **33**, 1528 (1994).
- J. Li, J. W. Mayer, and E. G. Colgan, *J. Appl. Phys.*, **70**, 2820 (1991).
- M. O'Reilly, X. Jiang, J. T. Beechinor, S. Lynch, C. Ni Dheasuna, J. C. Patterson, and G. M. Crean, *Appl. Surf. Sci.*, **91**, 152 (1995).
- H. K. Liou, J. S. Huang, and K. N. Tu, *J. Appl. Phys.*, **77**, 5443 (1995).
- A. J. Griffin, Jr., S. E. Hernández, and F. K. Brotzen, *J. Electrochem. Soc.*, **141**, 807 (1994).
- C.-K. Hu, B. Luther, F. B. Kaufman, J. Hummel, C. Uzoh, and D. J. Pearson, *Thin Solid Films*, **262**, 84 (1995).
- W. A. Lanford, P. J. Ding, W. Wang, S. Hymes, and S. P. Murarka, *Thin Solid Films*, **262**, 234 (1995).
- P. J. Ding, W. Wang, W. A. Lanford, S. Hymes, and S. P. Murarka, *Appl. Phys. Lett.*, **65**, 1778 (1994).
- J. Li, J. W. Mayer, Y. Shacham-Diamand, and E. G. Colgan, *Appl. Phys. Lett.*, **60**, 2983 (1992).
- H. Itow, Y. Nakasaki, G. Minamihaba, K. Suguro, and H. Okano, *Appl. Phys. Lett.*, **63**, 934 (1993).
- D. S. Gardner, J. Onuki, K. Kudoo, Y. Misawa, and Q. T. Vu, *Thin Solid Films*, **262**, 104 (1995).
- Y. Igarashi, T. Yamanobe, T. Yamaji, S. Nishikawa, and T. Ito, *Jpn. J. Appl. Phys.*, Part 1, **33**, 462 (1994).
- S. Hymes, S. P. Murarka, C. Shepard, and W. A. Lanford, *J. Appl. Phys.*, **71**, 4623 (1992).
- P. J. Ding, W. A. Lanford, S. Hymes, and S. P. Murarka, *J. Appl. Phys.*, **74**, 1331 (1993).
- J.-C. Chuang and M.-C. Chen, *J. Electrochem. Soc.*, **145**, 3170 (1998).
- J.-C. Chuang and M.-C. Chen, *Thin Solid Films*, **322**, 213 (1998).
- D. D. Wu, Masters Thesis, National Chiao-Tung University, Hsinchu, Taiwan (1996).
- Metallization in Silicon Processing for the VLSI Era*, Vol. 1, S. Wolf and R. N. Tauber, Editors, Lattice Press, Sunset Beach, CA (1986).
- K. Holloway, P. M. Fryer, C. Cabral, Jr., J. M. E. Harper, P. J. Bailey, and K. H. Kelleher, *J. Appl. Phys.*, **71**, 5433 (1992).