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Weiyi Li, Dongfei Pei, Daniel Benjamin, and Jen-Yung Chang

Plasma Processing and Technology Laboratory, Department of Electrical and Computer Engineering, University of Wisconsin-Madison, Madison, Wisconsin 53706

Sean W. King

Intel Corporation, Hillsboro, Oregon 97124

Qinghuang Lin

IBM T.J. Watson Research Center, Yorktown Heights, New York 10598

J. Leon Shohet^{a)}

Plasma Processing and Technology Laboratory, Department of Electrical and Computer Engineering, University of Wisconsin-Madison, Madison, Wisconsin 53706

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In this work, the authors report an investigation of the effects of cesium (Cs) ion implantation on both porogen-embedded and ultraviolet (UV)-cured (porous) SiCOH films. For porogen-embedded SiCOH, it was found that Cs ion implantation can greatly improve the elastic modulus. It can also increase the time-zero dielectric breakdown (TZDB) strength. It also leads to an increase in the k-value for medium and high Cs doses, but for low Cs doses, the k-value decreased compared with its pristine counterpart. For UV-cured SiCOH, it was found that Cs-ion implantation does not improve the elastic modulus. It also leads to lower TZDB field strength and much higher k-values than its pristine counterpart. These effects can be understood by examining the changes in chemical bonds. This treatment is shown to have the potential to help solve the problem of the demand for lower k-values and the concomitant weak mechanical strength of SiCOH. © 2017 American Vacuum Society. [http://dx.doi.org/10.1116/1.5001573]

I. INTRODUCTION

Being a part of solving the huge puzzle that extends Moore's law,¹ low dielectric-constant (low-*k*-value, or low-*k*) materials are being used to replace SiO₂ in back-end-of-line (BEOL) interconnections.² It is important to reduce the resistance-capacitance delay of the interconnects, which increases as the critical-dimension scaling proceeds downward.² Among some candidates for low-k interconnects, 3-8 organosilicate glass, which is often called SiCOH, SiOC, SiOCH, or SiOC:H, has been adopted for BEOL applications.^{9–11} In order to push the k-value of these materials even lower to meet the requirements of the International Technology Roadmap of Semiconductor,¹² porosities were introduced into these materials.^{13–18} The currently dominant method to introduce porosity is by embedding a sacrificial porogen material during SiCOH film deposition, and then removal the porogen with a combination of thermal heating and ultraviolet (UV) light exposure.14,19-23 This combined treatment is usually referred to as UV-curing.14,19-23 A UV-cured SiCOH film is porous and thus has a lower k-value than its nonporous counterpart.

However, a problem with porous SiCOH (p-SiCOH) is that its mechanical properties, e.g., elastic modulus and hardness, can be greatly weakened because of the porosity. For example, the elastic modulus of p-SiCOH can be 4–6 GPa while its nonporous counterpart shows values of 15–20 GPa.^{24–26} For comparison, the elastic modulus of SiO₂ lies in the range of 57–92 GPa.²⁷ These weak mechanical properties may cause problems during chemical–mechanical polishing and packaging.²⁸ A number of ways have been suggested to solve this problem.^{29–31}

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Our previous work³² showed that cesium (Cs) ion implantation in nonporous (no porogen involved) SiCOH low-kdielectric films can help improve its mechanical properties, including both the elastic modulus and hardness. The timezero dielectric breakdown (TZDB) characteristics were also improved. However, along with these improvements, the Csimplanted material exhibited an increase in its k-value and loss of hydrophobicity.

In this work, we investigate the effects of Cs-ion implantation on the mechanical and electrical properties of p-SiCOH dielectrics. In addition, the effects of thermal annealing after ion implantation are also investigated. We hypothesize that Cs-implantation changes to the physics and electrical properties are as a result of the implantation process. The details of sample selection, Cs ion implantation specifications, and sample characterization methods will be discussed in Sec. II. The measured data will be shown and discussed in Sec. III.

II. EXPERIMENT

Two types of SiCOH films, named V1 and V2, are used in this work. They were provided by Intel. The properties of them are listed in Table I.

Note that V1 has embedded porogens, and V2 is UV-cured V1.

The Cs ion-implantation conditions chosen for these samples are listed in Table II. These conditions were chosen

^{a)}Electronic mail: shohet@engr.wisc.edu

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TABLE I. Properties of SiCOH films used in this work.

Film type	UV cured?	Thickness (nm)	k-value	Porosity
V1	No	226.8	2.8	Porogen embedded
V2	Yes	181.5	2.1	42%

based on a stopping and range of ions in matter (SRIM) code simulation,³³ with the goal of keeping the Cs-ion distribution centered roughly in the middle of the film thickness. Because of the porosity of the V2 film, Cs ions with the same kinetic energy can travel deeper into V2 film than V1 film. The effective concentration, taking into account porosity, increases for the same dose, compared with the V1 film. Therefore, the ion energy and dose for the V2 film need to be reduced accordingly to achieve the same conditions as those for the V1 film.

After implantation, all the samples were cut into two pieces, one of which was not processed further. These samples were labeled as implanted only. The other piece was annealed in a vacuum heater at $300 \,^{\circ}$ C for 1 h and labeled as implanted and annealed. For comparison, samples of the *unimplanted* pristine films were also annealed in the same manner. Thus, there are four types of samples after this point: (1) pristine (P), (2) thermal-annealed only (A), (3) ion-implanted only (I), and (4) implanted-and-annealed (I-A).

The thickness of each sample was measured with a Rudolph AutoElII-VIS-3 ellipsometer. The ellipsometerbeam incident angle was fixed at 70° . Three wavelengths, 632.8 nm (R), 546.1 nm (G), and 405 nm (B), were selected for the incident beam.

The elastic modulus of each sample was measured using nanoindentation with a Hysitron TI 950 Triboindenter. A conventional Berkovich tip was used for these measurements. A comparison between measured effective moduli with a theoretical model developed by Stone^{34} allows us to determine the elastic moduli for the low-*k* film itself.^{26,32}

The *k*-values of the films were determined from their C-V characteristics.³⁵ The setup for the C-V measurement utilizes a Signatone H-150W probe station with a $5-\mu$ m diameter pin probe, an HP 4285A LCR meter, and a metal–insulator–semiconductor (MIS) stack structure. The metal dots of the MIS structure are made with a silver (200 nm)/titanium (100 nm) bilayer, deposited on the SiCOH films with vacuum evaporation. Silver was deposited on top of the titanium. The area of the metal dots is 4×10^{-4} cm² on the average, as measured with an optical microscope. The C-V characteristics were measured at 100 kHz on at least 12 separate metal dots with one measurement on each dot to reduce the system error by averaging the measured capacitance. The *k*-value was calculated using the following expression:

$$k = \frac{t \cdot C_{\text{accumulation}}}{A \cdot \varepsilon_0}$$

where k is the k-value, t is the film thickness, $C_{\text{accumulation}}$ is the averaged capacitance in the accumulation region, A is the metal dot area, and ε_0 is the vacuum permittivity. Because the conducting area on the bottom of the sample is much larger than the top electrode, the calculated k is usually over-estimated (since A is under-estimated). Therefore, the k-values of the pristine films supplied by Intel were used to calibrate our calculated k-values.

The electric field for TZDB was determined from the I-V characteristics, utilizing a setup similar to that used for the C-V characteristics, except that the LCR meter was replaced with a Keithley 6487 picoammeter. A voltage ramping rate of approximately 1 V/s was used. The criterion for TZDB is defined as the point at which the current density exceeds 0.5 A/cm².

Chemical-bonding information was obtained using Fouriertransform infra-red (FTIR) spectroscopy measurements, performed with a Thermo Micro FTIR Spectrometer. The measured spectra were manually baseline-corrected one-by-one

TABLE II. Sample labels, preparation conditions, measured elastic moduli, and k-values. P, pristine; I, implanted; L, low dose; M, medium dose; H, high dose; and A, annealed.

Film type	Sample label	Cs ion energy (keV)	Cs ion dose (cm^{-2})	Annealing (vacuum)	Elastic modulus (GPa)	k-value
V1	V1-P	NA	NA	NA	10	2.81
	V1-I-L	100	1×10^{13}	NA	15	2.76
	V1-I-M	100	4×10^{13}	NA	16	2.90
	V1-I-H	100	8×10^{13}	NA	20	3.23
	V1-A	NA	NA	300 °C, 1 h	11	2.81
	V1-I-L-A	100	1×10^{13}	300 °C, 1 h	9	2.66
	V1-I-M-A	100	4×10^{13}	300 °C, 1 h	9.5	2.97
	V1-I-H-A	100	8×10^{13}	300 °C, 1 h	9.5	2.97
V2	V2-P	NA	NA	NA	10	2.1
	V2-I-L	50	6×10^{12}	NA	9	2.13
	V2-I-M	50	$2.4 imes 10^{13}$	NA	10	2.44
	V2-I-H	50	$4.8 imes 10^{13}$	NA	10	3.11
	V2-A	NA	NA	300 °C, 1 h	5	2.19
	V2-I-L-A	50	6×10^{12}	300 °C, 1 h	5.5	2.20
	V2-I-M-A	50	$2.4 imes 10^{13}$	300 °C, 1 h	6	2.54
	V2-I-H-A	50	4.8×10^{13}	300 °C, 1 h	6	2.71

(i.e., based not on a generic baseline, but an individual baseline for each curve). The curves were not normalized with respect to the film thickness in order to estimate the absolute number of bonds instead of their relative concentration.

Elemental composition information was found with XPS depth-profiling, as measured with a Thermo K-Alpha XPS. Only Si, C, and O could be detected with this technique. H and Cs could not be detected with XPS, because H has no core electron and Cs has too low a concentration in these films to be distinguishable from noise.

III. RESULTS AND DISCUSSION

This section is divided into three parts. Section III A covers the effects of Cs implantation on porogen-embedded SiCOH, Sec. III B describes the effects of Cs implantation on porous SiCOH, and Sec. III C discusses the effects of argon implantation on porogen-embedded SiCOH (as in Sec. III A). The latter section is introduced for the purpose of indicating whether the cesium implantation might be a chemical or physical effect.

A. Effects of Cs implantation and annealing on porogen-embedded SiCOH (V1 type)

1. Film thickness

The measured film thicknesses were expressed as a relative shrinkage (in percent) with respect to the corresponding pristine-film thickness and are shown in Fig. 1. Note that each line in Fig. 1 represents a particular processing sequence. Several effects can be observed from Fig. 1. First, "300 °C, 1 h" thermal annealing causes only a very small shrinkage, about 1%, of both the pristine V1 film. This indicates that the chosen annealing condition is moderate so that it does not cause film densification because the thermal budget is low enough so avoid significant chemical changes to the pristine film.

Second, for implanted-only samples, the Cs implantation causes the film thickness to shrink more significantly, and the shrinkage increases as the Cs dose increases.



Fig. 1. (Color online) Film thickness shrinkage, relative to the pristine film thickness, measured by ellipsometry.

Third, the implanted-and-annealed samples show an additional but slight shrinkage on top of the shrinkage caused by implantation alone. The magnitude of this additional shrinkage is similar to the magnitude of shrinkage of the annealedonly samples (comparing both to their corresponding pristine samples). This indicates that the mechanisms for these two effects, one due to Cs implantation and one due to thermal annealing, can be separated.

2. Elastic modulus

The elastic moduli for various samples, obtained from nanoindentation, are listed in Table II. Due to the relatively thin thickness of the films, the directly measured effective moduli from nanoindentation are much higher than the actual elastic moduli of the films themselves. This is because of the contribution from the Si substrate to the nanoindentation measurements.^{30–32} In order to extract the actual elastic moduli, a theoretical model developed by Stone,³⁴ which includes the contribution of the silicon substrate into account when analyzing the measured data, was used. The method used to interpret this type of data has been introduced elsewhere.^{26,32} Even so, it is still advised that the values of elastic moduli listed in Table II may have only a relative meaning.

It can be seen from Table II that the uncured V1 film is greatly strengthened by Cs implantation, which improves even more as the Cs dose increases. UV curing, on the other hand, does not show such improvement (compare V1-P with V2-P). Especially for the high dose case, the elastic modulus increases from about 10 GPa for the V1-P sample to about 20 GPa for the V1-I-H sample, a 100% improvement. Conversely, annealing *after* implantation caused no improvement but a decrease in the modulus of the Cs-implanted samples.

In some recent work, the improvement of the mechanical properties was attributed to an increase in the Si-O-Si network structure.^{22,29,36} We observe similar effects here. Gaussian peak deconvolution fitting³⁷ was used to separate the contributions to the FTIR spectra from the Si-O-Si cage structure, network structure, and suboxide structure.^{9,37} An example of this deconvolution, performed on sample V1-P, is shown in Fig. 2. The inset table in the figure shows fitted parameters for three distinct Gaussian peaks (cage, network, and suboxide). Each peak has four parameters. These are: y_0 —vertical base offset, x_c —horizontal center, w—peak width (twice sigma, not FWHM), and A-integrated area under the curve. In this work, the values of fitted x_c and A are essential, because x_c indicates which of the three Si-O-Si structures a given peak represents, and A indicates what the proportion of this particular structure out of the total three is. After fitting, the peaks can be labeled as cage, network, and suboxide according to their corresponding x_c values. The A values of each peak can then be used to calculate an area ratio using the following expression:

$$\frac{A_{\text{network}}}{A_{\text{total}}} = \frac{A_{\text{network}}}{A_{\text{cage}} + A_{\text{network}} + A_{\text{suboxide}}}$$

For each sample, there is such an area ratio, which can be used as a measure of its mechanical properties.



FIG. 2. (Color online) Example of FTIR spectral deconvolution of the Si-O-Si structure.

The area ratios for the various samples are shown in Fig. 3. It can be seen that the area ratio increases after Cs implantation, and it increases further when the Cs dose increases. On the other hand, thermal annealing, either by itself or after Cs implantation, shows a decrease of the area ratio. UV curing by itself also decreases the area ratio. These findings are consistent with the measured elastic moduli in Table II. That is, the increase of the modulus is consistent with the increase in the network structure.

3. k-values

The normalized *k*-values for the various samples are shown in Table II. The relative changes of the *k*-values of the annealed only, implanted only, and implanted-and-annealed samples with respect to their corresponding pristine samples are shown in Fig. 4.

From both Table II and Fig. 4(a), several observations about the *k*-value changes of the V1 films can be made. First, the *k*-value of the V1 film is almost not affected from



FIG. 3. (Color online) FTIR area-to-area ratio $A_{network}/A_{total}$ of various samples. P, pristine; I, implanted; L, low dose; M, medium dose; H, high dose; and A, annealed.



FIG. 4. (Color online) Relative changes in percentage of the *k*-value of processed (a) V1, and (b) V2 samples with respect to their corresponding pristine sample's *k*-value.

thermal annealing alone. Second, Cs-implanted samples do not always show an increased *k*-value, which is different from the results previously presented for nonporous SiCOH, in which Cs implantation caused an increase in the *k*-value for the three chosen doses.³² For the V1 samples, low-dose Cs implantation actually lowers its *k*-value slightly. For medium and high doses, the *k*-values of all samples increase with Cs dose. In addition, the porogen-embedded V1 films show a much smaller increase in the *k*-value compared with the UV-cured V2 film. This is similar to what was found in our nonporous SiCOH work.³²

Third, in many cases, the percentage of the *k*-value increase is less for the implanted-and-annealed samples compared with their corresponding implanted-only samples. This is the opposite effect compared with what was observed in our nonporous SiCOH work,³² in which thermal annealing always caused the *k*-value to increase.

To explain this effect for nonporous SiCOH,³² it was stated that the samples were undesirably oxidized during furnace annealing, due to the presence of oxygen in the atmosphere contacting the still-hot samples after annealing. In this work, however, no such oxidation occurred because a vacuum heater was used to perform the annealing instead of the furnace. Thus, the samples can be sufficiently cooled down to room temperature before making contact with the atmosphere. It appears that the vacuum annealing seems to be able to partially mitigate the k-value increase caused by Cs implantation, especially for medium and high Cs doses. This might be due to two reasons. One, some porogen can be removed after annealing, leaving pores behind that reduce the k-value. Two, water which was absorbed after Cs implantation can be driven out during vacuum annealing, which also helps to reduce the k-value.

In order to determine the mechanism causing the observed change to the *k*-values, the FTIR spectra of the V1 series samples (Pristine, implanted only, and implanted-and-annealed), are shown in Fig. 5. For comparison, the FTIR spectrum of the UV-cured V2-P (Pristine) sample is also shown in Fig. 5. It can be seen that for both the implanted-only and the implanted-and-annealed samples, there is a clear carbon loss,³⁷ which increases as the Cs dose increases. This explains the increase of the *k*-values.

On the other hand, compared with the *k*-value increase for either low, medium, or high Cs doses in our nonporous SiCOH work,³² the *k*-value increase for porous SiCOH for a given dose is suppressed in the porogen-embedded V1 film. This can be attributed to pore creation in the film from the implanted Cs ions. This is especially likely near the surface of the film.

One piece of evidence for this is shown in the FTIR spectra in Fig. 6. In Fig. 6 it can be seen that the $-CH_x$ peaks of the implanted-only and implanted-and-annealed samples decrease, and this decrease is similar in magnitude to that produced by UV curing, i.e., the difference between the V1-P



FIG. 5. (Color online) FTIR spectra of the SiC– H_3 bond and the Si– CH_3 bond of the V1 I and I-A samples compared with pristine V1 (uncured) and V2 (UV-cured) samples. P, pristine; I, implanted; L, low dose; M, medium dose; H, high dose; and A, annealed.



FIG. 6. (Color online) FTIR spectra of $-CH_x$ (porogen-related) peaks of V1 I and IA samples compared with pristine V1 (uncured) and V2 (UV-cured) samples. P, pristine; I, implanted; L, low dose; M, medium dose; H, high dose; and A, annealed.

and the V2-P samples. With a higher Cs dose, the peaks decrease even more. Thermal annealing also leads to a decrease of these peaks, which supports our aforementioned explanation that annealing can remove the porogen. It is widely accepted that these $-CH_x$ peaks are indicators of the porogen amount left inside the film.^{22,36} Therefore, from Fig. 6, we deduce that the implanted Cs ions, either by themselves or combined with subsequent annealing, have resulted in partial removal of the porogen that was embedded in the pristine film, resulting in pore creation. This effect explains the suppression of the *k*-value increase, compared with the nonporous SiCOH case.³²

Additional support for this hypothesis can be found from the XPS depth profile. Figure 7 shows the XPS depth profiles of pristine V1 and V2 samples, as well as that for Csimplanted V1 samples. It can be seen from Fig. 7, similarly to what was found in the nonporous SiCOH work,³² that carbon loss within approximately the top 40%–60% of the film near the surface can be observed. The degree of this loss is higher as the Cs dose increases. In addition, a corresponding increase in oxygen can be observed in the same region.

However, there is a different phenomenon at work in the case of porous SiCOH which must be considered. It is related to silicon. Unlike what was found for nonporous SiCOH,³² where the silicon composition was almost unchanged by Cs implantation, in Fig. 7(a) it can be seen that the Si composition in the aforementioned top 40%–60% of the film thickness region increases. In addition, the degree of this increase is stronger as the Cs dose increases. Since Si atoms are believed to be more difficult to remove from the film, we deduce that this change in composition is a reflection of losing porogen material, which, to the best of our knowledge, contains carbon, oxygen, and hydrogen, but not silicon. Note that XPS measurements do not give the absolute but only the relative elemental concentrations, thus removing other elements than silicon results in an increase of the relative concentration of



FIG. 7. (Color online) XPS depth profiles of (a) silicon, (b) carbon, and (c) oxygen atoms of V1 series of samples. Pristine V2 sample is also shown for comparison. Horizontal axis is percentage of depth, normalized with respect to individual film thickness of each sample. P, pristine; I, implanted; L, low dose; M, medium dose; H, high dose; and A, annealed.

silicon. Therefore, this observation supports our hypothesis of porogen removal due to Cs implantation.

4. TZDB field strength

Figure 8 shows box-plots of the TZDB electric-field statistics for all the samples, measured with I-V characteristics at room temperature. The box-plot statistics are based on at least 12 measurements on separate metal dots, one measurement on each dot. Several observations can be found from Fig. 8. First, the annealed-only samples do not show any improvement (increase) in the TZDB field as compared with their corresponding pristine samples. In fact, for both V1 and V2 films, the annealed-only samples show *lower* TZDB fields.

Second, for uncured V1 films, the medium and high-dose Cs implantations in fact increase the TZDB field, while the low-dose implantations do not show this effect. Third, annealing after Cs implantation in general either improves or



FIG. 8. (Color online) Box-plots of the TZDB electric field of (a) V1, and (b) V2 samples. Labels: P, pristine; I, implanted; L, low dose; M, medium dose; H, high dose; A, annealed.

holds the TZDB field compared with that of the corresponding implanted-only samples.

The improvement of TZDB due to Cs implantation may be explained by film densification, which was observed with the thickness measurements. In fact, some Cs-implanted uncured samples show equal or better TZDB fields compared with the corresponding UV-cured samples. One example is V1-I-H-A compared with V2-P. The additional improvement due to subsequent annealing can be explained by considering the effect of high-temperature-enhanced diffusion of the implanted Cs ions, which redistributes the Cs ions more uniformly, thus decreasing the local electric-field peak.

B. Effects of Cs implantation on porous SiCOH (V2 type)

1. Film thickness

As can be seen from Fig. 1, in general for each of three Cs doses, the UV-cured V2 film shows a much smaller shrinkage compared with the uncured V1 film. This difference indicates that the thickness shrinkage may be primarily due to the presence of the porogen in the V1 film, and a corresponding lack of its presence in the V2 film.

2. Elastic modulus

As can be seen from Table II, for the UV cured V2 film, no improvement in the elastic modulus was observed, either after Cs implantation or implantation followed by annealing. In all implanted-only or implanted-and-annealed cases, the elastic moduli are either equal or lower than that of V2-P sample. This is similar to our previous findings,³² i.e., there is no benefit for the elastic modulus with Cs implantation in an already UV-cured film.

3. k-values

As can be seen from both Table II and Fig. 4(b), implanting Cs into UV-cured SiCOH caused a significant increase in the k-value, especially for the medium and high-dose samples. Unlike the low-dose case for the V1 film, the V2 film showed no case where the k-value decreased. This is consistent with the aforementioned explanation for the V1 film that implanting Cs leads to partial removal of the porogens. Presumably, this does not exist in the V2 film. The implanted and annealed V2 samples also show increased k-values.

4. TZDB field strength

As can be seen from Fig. 8(b), compared with the pristine V2-P sample, all the implanted-only and implanted-andannealed samples show a lower TZDB field strength. This decrease is more significant when the Cs dose is higher.

C. Effects of Ar implantation on porogen embedded SiCOH (V1 type)

Finally, in order to answer questions as to whether the Cs ions cause a chemical effect or physical effect during the implantation, and whether other ions can exhibit the same effect, argon (Ar) was implanted into the V1 films. Because of the inert nature of Ar, it is assumed that the implanted Ar ions can only have a physical bombardment effect, but no chemical effects on the V1 films. The results of elastic modulus and k-value measurements are summarized in Table III. Two kinetic energies, 100 and 50 keV, were chosen. These energy values chosen were (for 100 keV) to give the Ar ions the same kinetic energy used for the Cs ions in Table II, or (for 50 keV) to keep the implantation depth similar to that found for the 100 keV Cs-ion implantation, based on our SRIM code simulation. In both cases, the Ar doses were set to be the same as the Cs doses.

It can be seen from Table III that 100-keV Ar implantation increases the elastic modulus of the V1 samples from 11 to 15 GPa, which is smaller when compared with the effects of Cs implantation (see Table II). The 50-keV Ar implantation, on the other hand, had almost no effect on the elastic modulus.

From the measured k-values shown in Table III, it can be seen that Ar implantation only increases the k-value under all conditions. There is no case where the k-value decreases, as does the low-dose Cs implantation of the V1 films. The relative increase of the k-value is larger than the case for Cs implantation, as can be seen by comparing Table III with Fig. 4(a). The two entries "??" in Table III indicate that the Ar implantation resulted in tens of volts of flat-band voltage shift so that the accumulation region for these samples could not be found.

Therefore, it can be inferred that either (1) Cs implantation is not a purely physical effect, i.e., the implanted Cs ion breaks chemical bonds not only because of its kinetic energy, but also because of its chemical properties, or (2) what Cs implantation causes is a purely physical effect, similar to Ar implantation, but the higher momentum of the Cs ion makes the difference. Note that when the kinetic energy is kept the same, an ion's momentum is proportional to the square root of the ion's mass, i.e., $p = \sqrt{2mE}$.

IV. CONCLUSION

In conclusion, Cs implantation in porogen-embedded SiCOH samples exhibits benefits such as improving the elastic modulus, improving TZDB, and partially removing porogen to make pores. These effects are all beneficial for the requirements for low-*k* dielectrics. In addition, the undesirable *k*-value increase can be suppressed to a relatively low

TABLE III. Sample labels, preparation conditions, measured elastic moduli, and k-values.

Film type	Ar ion energy (keV)	Ar ion dose (cm^{-2})	Elastic modulus (GPa)	k-value	Relative change of <i>k</i> -value (%)
V1	NA	NA	11	2.81	+0
	100	1×10^{13}	12	3.09	+10.0
	100	4×10^{13}	12	??	NA
	100	$8 imes 10^{13}$	15	??	NA
	50	1×10^{13}	11	3.06	+8.90
	50	4×10^{13}	11	3.22	+14.6
	50	8×10^{13}	12	3.54	+26.0

value compared with the increase in the case of nonporous SiCOH. In fact, if the Cs dose is low enough, the k value decreases. This phenomenon suggests that Cs implantation of porogen-embedded SiCOH may be more useful than implantation of nonporous SiCOH. However, implanting Cs into UV-cured p-SiCOH leads to an unacceptable increase of the k-value, and it does not increase the elastic modulus. For comparison, Ar implantation in porogen-embedded SiCOH leads to some increase in the elastic moduli but shows a larger increase in the k-values, the latter deviating from the requirements for low-k dielectrics. No Ar implantation condition that led to a net decrease in the k-value was found.

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