Defect profiling in FEFET Si:HfO₂ layers

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ABSTRACT

Ferroelectric Si doped HfO₂ is a promising candidate for future generation memory devices. However, such devices are vulnerable to significant threshold voltage shifts, resulting from charge trapping in oxide defects. We use complementary characterisation and modelling techniques to reveal significant electron trapping/de-trapping behaviour, together with strong temperature dependence of the electron emission kinetics in ferroelectric layers, which results from the onset of polarisation of the ferroelectric layer. This can lead to an apparent difference in the defect characteristics in ferroelectric-HfO₂ than the paraelectric-HfO₂ structures they are shown to closely resemble when this contribution is decoupled. Results demonstrate the presence of a defect band closely aligned to the silicon conduction band, which can easily be accessed during device operation.

Continuation of the trends in non-volatile memory device performance will require further reductions in power consumption, combined with increases in device endurance and access time. However, a roadblock is being reached in that the technologies currently being applied are approaching their scaling limits. For this reason, technologies are being explored to enable a continuation of performance scaling, albeit incorporating alternative materials and device concepts. Controlled doping of HfO₂ (with, for example, Al, Si, Zr), combined with an optimum thermal budget, enable the presence of the orthorhombic phase displaying ferroelectric characteristics [1,2]. The integration of this ferroelectric phase of HfO₂ into a MOSFET transistor gate dielectric offers the potential to continue this memory device dimension scaling, when combined with the added benefits of CMOS compatibility (enabling synergy with parallel logic device development), provide a very promising candidate for future generation emerging memory technologies.

Ferroelectric behaviour in silicon doped HfO₂ thin films has been reported for films of thickness in excess of 5 nm [3]. At such physical thickness, the presence (and absolute number) of defects give rise to capture and emission of electrons injected in the FE-layer, which also modulate the device performance. The combination of this defect-dominated behaviour together with the ferroelectric polarisation presents quite a challenge for device operation and reliability engineering [4] and is the objective of this work. We elucidate the energy distributions of oxide defect states in Si doped-HfO₂ for FE-RAM applications using multiple defect modelling techniques.

Trapping and noise measurements were performed on 9.5nm Si-doped HfO₂/TiN [4] devices, where a thin oxide interfacial layer was present between the silicon substrate and ferroelectric layers, whereas a thicker (7.5 nm) oxide layer was used in the photo-depopulation study. Results are benchmarked to undoped HfO₂/TiN stacks subjected to

similar thermal budget on planar transistor structures, which display ferroelectric and paraelectric behaviour, respectively. Both stacks show significant trapping behaviour, with the complexity of the threshold voltage (V_t) reduction due to polarisation of the ferroelectric layer.

The ferroelectric nature of the silicon doped HfO_2 is demonstrated in PV loops shown in Fig. 1A, with $\sim 4\text{V}$ coercive voltage ($2V_c$) and $\sim 20\text{ }\mu\text{C}/\text{cm}^2$ remnant polarisation ($2P_r$), whereas the paraelectric nature of the undoped HfO_2 is evident, independent of the number of cycles applied, unlike the FEFET case, where the coercive voltage reduces with cycling (most notable in the positive direction). Analysis of capacitance measurements of layers of different HfO_2 and Si:HfO₂ thickness results yield an effective permittivity of ~ 30 for (ferroelectric) Si:HfO₂ higher than that of HfO_2 (~ 20). These are derived from the plot of physical thickness versus equivalent oxide thickness (EOT), shown in Fig. 1B. The y-intercept of this plot suggests a thinner interface layer in the case of the FE-HfO₂, 0.5 nm instead of 0.8nm for undoped HfO_2 (Fig. 1B), details which are utilised in the modelling works described below.

FIG. 1. (A) Polarisation-Voltage traces for Si:HfO₂ and HfO₂ stacks, showing ferroelectric and paraelectric behaviour, respectively, and (B) Equivalent oxide thickness values, derived from Capacitance measurements, versus physical thickness show higher- κ_{eff} and a lower interface layer thickness in FEFET.

Conventional extended Measure-Stress-Measure (eMSM) was used for charge trapping studies [5], where the drain current is monitored during both the stress (at $V_g = V_{\text{stress}}$) and intermittent relaxation phases, (at $V_g = V_t$ prior to stress, or V_{t0}), as depicted in Fig 2A. Using the initial I_d - V_g trace, the current during each of the relaxation phases is transformed into a threshold voltage shift. The planar transistors studied in this work were $10 \times 1\text{ }\mu\text{m}^2$, and a fresh device was used at each stress condition: no pre-cycling or wake-up was performed. The threshold voltage shift (ΔV_t) induced by stressing, together with the relaxation thereof during relaxation is shown in Fig. 2B and C. Devices are compared after stressing at constant effective electric field (E_{eff}), which is calculated as shown in (1)

$$E_{\text{eff}} = \frac{(V_{g \text{ stress}} - V_{t0})}{EOT} \quad (1)$$

Significant charge trapping during stress, together with de-trapping during relaxation (over 8 decades of time) are observed, for HfO₂ and FE-HfO₂ stacks (Fig. 2B, and C respectively). Interestingly, limited de-trapping is observed from the FE-HfO₂ if $t_{\text{relax}} < 100\text{ }\mu\text{s}$, and relaxation continues to a negative ΔV_t value i.e. lower than the initial device threshold voltage. The relaxation of the ferroelectric layer to a lower V_t than the starting point is interpreted as a (partial) change of the polarisation state during the eMSM sequence, as discussed below.

FIG. 2. Contour plot showing ΔV_t as a function of stress and relaxation time, applied using (A) eMSM, measured at 25°C . The initial device threshold voltage is used as V_{t0} for (B) HfO₂ shows strong trapping/de-trapping and (C) FE-HfO₂, also shows relaxation to negative V_t after longer relaxation cycles.

This anomalous feature is not observed in HfO₂, which displays conventional positive bias temperature instability (PBTI) electron-trapping behaviour (i.e., $\Delta V_t > 0$ post-relaxation, across all bias conditions). Negligible temperature dependence of electron trapping and de-trapping is observed for HfO₂, which implies a large density of defects are readily accessible for trapping/de-trapping even at low bias conditions, shown in Fig. 3.

However, in the FEFET case a strong temperature dependence of de-trapping is observed (Fig. 3), where the rate of ΔV_t reduction considerably increases at lower temperatures (and shorter stress times (not shown)). It is seen that for the lower temperatures, V_t can relax to a lower value than the initial case (i.e. $\Delta V_t < 0$). This is believed to result from the commencement of the positive (stress and relaxation) voltage-induced polarisation of the FE layer to the low V_t state, and hence the lower V_t value. That this process can take place at these low voltages (for $V_{\text{stress}} \sim 1.5\text{V}$, $V_{\text{relax}} \sim 0.5\text{V}$), after long (stress and relaxation) times is consistent with results on the time and bias dependence of polarisation [4]. This effect is more pronounced at lower temperatures, also in agreement with a recent report suggesting increased orthorhombic phase in Si-doped HfO₂ on reducing temperature [6].

FIG. 3. ΔV_t versus relaxation time for HfO₂ and FEFET samples, as a function of temperature, showing significant difference in temperature activation of the de-trapping process in the FEFET sample from the HfO₂ reference, with *negative* ΔV_t values measured at -35°C after longer relaxation times.

The time constants for capture, τ_c , and emission, τ_e , of an electron into/from a given defect are described in (2)

$$\tau_c = \tau_0 \exp\left(\frac{E_{Ac}}{kT}\right) \quad \tau_e = \tau_0 \exp\left(\frac{E_{Ae}}{kT}\right) \quad (2)$$

where, τ_0 is an effective time constant that incorporates the carrier concentration, elastic tunnelling probability and trap capture cross section (weakly bias and temperature dependent), E_{Ac} and E_{Ae} are the effective activation energies for charge capture and emission, respectively, k is the Boltzmann constant, and T is absolute temperature [7]. In this method, the activation energy distributions (for E_{Ac} and E_{Ae}), expressed in (2), are optimized to obtain the effective capture and emission time constants, which in turn can consistently reproduce the measured ΔV_t stress/recovery behaviour over a wide temperature range upon integration over the experimental time range [7]. The capture and emission time (CET) map thus obtained, can further be utilized to study the behaviour of charge trapping kinetics, as well as identify the defect populations that are accessible under varying temperature conditions.

eMSM results collected at multiple temperatures (-40 to 125°C for HfO₂; -40 to ~80°C for Si:HfO₂) were used to obtain the optimized CET maps presented in Fig. 4. The activation energy distributions for charge capture (E_{Ac}) and emission (E_{Ae}) for an ensemble of defects in the gate-stack are modelled using two bivariate Gaussian sub-populations [8], R_1 and R_2 , which accurately reproduce PBTI results of the HfO₂ device, shown in Fig. 4A. Consequently, they may be associated with charging defects in the HfO₂ (R_1) and interface (R_2) layers and consistent with previous findings on logic transistor devices [7]. For the FEFET devices, only V_t values greater than zero were used, as this model does not

consider the ferroelectric layer contribution to threshold voltage. Nonetheless, valuable findings can be derived for such samples. The CET map calibrated to PBTI results of the FEFET devices reveals lower E_{Ac} and E_{Ae} for R_1 (Fig. 4B), indicating an impact of the onset of FE-HfO₂ layer polarisation, combined with differences in activation energies, on charge-trapping kinetics of the defects. This suggests a significant modification of the accessibility of the defect structure within the gate-stack upon incorporation of the orthorhombic phase resulting in an unfavourable activation energy distribution. This is also evident by faster discharging from oxide defects at lower temperatures observed in FEFET relaxation transients (Fig. 3).

FIG. 4. Capture and Emission activation energy maps generated from eMSM PBTI data at multiple temperatures for (A) HfO₂ and (B) Si:HfO₂ samples, stressed at ~3 MV/cm, show lower E_{Ac} and E_{Ae} for the R_1 (high- κ) defect in the FEFET sample. Each distribution (R_x ($x=1, 2$)) is a pdf of corresponding defect density and is normalized to its respective peak. Colour scale may be used to understand the relative height and distribution of R_x .

The defect charging was also studied using Trap Spectroscopy by Charge Injection and Sensing (TSCIS). In this technique, defects in HfO₂ and Si:HfO₂ are filled by direct tunnelling from the channel (through the thin interface layer), the depth of which is related to stress time, while their energy location is related to stress voltage, and the resultant band bending [9]. Emission behaviour is not considered in this technique, so short relaxation times (<10 ms) are applied to the eMSM-like sequence at 25°C. The calculated defect depth/energy profiles are plotted in Fig. 5, where it is seen that the FE-HfO₂ defect band is located at lower energy, than the HfO₂ case: 1.85 eV (0.25 eV) for FE-HfO₂, and 1.7 eV (~0.4 eV) for HfO₂, with respect to the HfO₂ (Silicon) conduction band edge. This seems to agree with lower E_{Ac} and E_{Ae} in the FE-HfO₂ CET map (Fig. 4).

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FIG. 5. Defect profiles generated from TSCIS for (A) HfO₂ and (B) Si-doped HfO₂ showing an electron trap defect band, located at ~150 meV lower defect band energy for the FEFET sample, measured at 25°C.

During the stress phases of an eMSM sequence, the stress voltage remains unchanged, but as the FEFET V_i can become lower than the starting point (depending on V_{stress} , t_{stress} and t_{relax}), this implies an increase in effective field during measurement of some devices, and increasingly so at lower temperatures. This is believed to play a role in the lower E_{Ac} and E_{Ae} values for the FEFET sample shown in Fig. 4, and defect band energy level in Fig. 5, both of which utilised data collected at low temperatures. This introduces a significant challenge in efforts to model the stress induced device degradation, and indeed opens questions on the learning that can be derived from such analysis. However, referring to Fig 3 again, it is noted that the emission kinetics of the highest temperature FE-HfO₂ dataset mirror those of HfO₂, which suggests the dominance of the conventional trapping effects at this temperature. This is used as an approach to decouple the ferroelectric polarisation effects, which increasingly plays a role at lower temperatures, from charge trapping, which dominates at higher temperatures, and thereby enables profiling of the intrinsic defect levels in FEFET's.

The data presented thusfar studied the impact of stress and relaxation time on resultant changes in V_i . Within this dataset, information on both capture and emission processes are collected, as illustrated for one bias condition in Fig 4. Extending this dataset to different stress bias conditions, enables modelling with the Comphy tool, which uses both 2-state non-radiative multi-phonon and double-well models, to account for charging of pre-existing oxide defects, and defect creation believed to be related to stress-induced hydrogen relocation, respectively [10]. As for other samples with thin interface layer, and high- κ layer, these systems are modelled as bilayer stacks, with a single shallow defect band, enabling electron capture during charging, and emission during subsequent discharging or relaxation phases. The simulations reveal information on each of the defect bands: location, width ($\langle E_t \rangle \pm \sigma_{E_t}$), density of defects therein (N_{ot}), and two parameters (R and S) which describe the distributed nature of the energy barriers to be overcome for transition between the two charge states [10].

Data collected at temperatures between -30°C and 125°C were used for the HfO₂ sample. The measured ΔV_i during a

single eMSM trace, together with those simulated by Comphy, are plotted in Fig. 6A for a range of bias conditions. The capture and emission processes are well described by the Comphy simulations, using the fit parameters in Fig. 6C for two defect bands, located in the interface and high- κ layers, respectively. The defect band in HfO₂, ~2 eV below the conduction band minimum ($E_{c \text{ min. HfO}_2}$), or ~0.35 eV above the silicon conduction band minimum ($E_{c \text{ min. Si}}$), is consistent with the low activation energy defect (labelled as R1), in Fig. 4, and increased charge trapping during PBTL measurements (Fig. 3). The interface layer defect band has lower density, higher energy, and S parameter (than in the HfO₂ layer), consistent with the higher E_{Ac} and E_{Ae} values for R₂ in Fig. 4. Bearing in mind the polarisation induced impact on trap emission discussed above, only the 80°C dataset was used for defect band modelling, as shown in Fig. 6B; where the interface layer defect parameters obtained for the undoped HfO₂ case were applied. The parameters of the FE-HfO₂ defect band, obtained in this manner, are surprisingly similar to those of HfO₂ (Fig. 6), and indeed the defect energy level and width are similar to those previously reported for ferroelectric HZO [11], in contrast to the marked differences observed from the respective CET maps, and TSCIS profiles (Figs. 4, 5).

FIG. 6. Measured ΔV_i -t data at varying effective electric field (symbols), during consecutive stress (incrementing from ~30ms to ~2000s) and relax (1ms-10s, in grey boxes) regimes for (A) HfO₂ and (B) FE-HfO₂ samples, at ~80°C. Solid lines are Comphy fits of measured dataset. Each stress and relax step are plotted on their own individual log scale. The ΔV_i reduction during relaxation is well described by Comphy simulations, using fit parameters for interface and high- κ defect bands shown below the respective panel.

The insertion of a 7.5 nm tunnel barrier between the silicon and FE-HfO₂ layer (in place of the thin interface layer) enables defect charging and subsequent discharge via controlled photon energy which can be captured electrically [12, 13] and transformed into an exhaustive photo depopulation spectrum (EPDS). The flat-band voltage calculated as a function of photon energy is plotted, pre- and post-defect charging in Fig. 7.

Clearly shown in the post-charged spectra for both samples are defect bands at ~2 eV below $E_{c \text{ min. HfO}_2}$, or ~0.3

eV above $E_{c, \min, Si}$, consistent with Comphy profiles (Fig. 5). Based on the inferred trap energies inside the HfO_2 bandgap, these states were tentatively associated with intrinsic polaronic and bi-polaronic states in the disordered regions universally encountered in deposited HfO_2 insulating films [12, 13].

FIG. 7. Plots of Flat band voltage, (V_{fb}), and Spectral Charge Density (SCD) versus photon energy for (A,B) HfO_2 and (C, D) FE- HfO_2 samples, showing prominent defect bands at ~2 and ~3 eV photon energy (or energy from the dielectric conduction band edge).

S_{V_g} spectra versus frequency show higher values for HfO_2 (Fig. 8) in agreement with measurements of charge trapping kinetics (Fig. 3) and Comphy calculations (inset Fig. 6) showing higher defect density in the HfO_2 sample. On reducing frequency, the FE- HfO_2 trend converges to that of undoped HfO_2 , suggesting an increase in defect density deeper in the FEFET's, consistent with the wider defect band seen in Comphy fit parameters. S_{V_g} (at 10 Hz) increases with V_{ov} , indicating presence of a defect band, again with higher density in HfO_2 .

FIG. 8. $1/f$ noise measurements as a function of (A) frequency and (B) V_{ov} for HfO_2 and FEFET samples measured at 25°C.

Thusfar, results from multiple techniques have shown a defect band in FE- HfO_2 layer at an energy level similar to

(Figs. 6 and 7), or lower than (Fig. 5) PE- HfO_2 . This apparent contradiction is also consistent with lower E_{Ac} and E_{Ae} values (Fig. 4) in the ferroelectric sample, but can be explained by the impact of temperature on the measurements, and in particular on the ferroelectric layer. When techniques which are not sensitive to this ferroelectric contribution (e.g. using thick interface layer, or at high temperature), the defect distribution evaluated, with energy level ~2 eV below $E_{c, \min, HfO_2}$ is believed to accurately describe the defect levels accessed with these positive bias stress conditions. However, more accessible shallow defect locations become apparent by the increased contribution of the ferroelectric contribution at lower temperatures, emphasising the challenge in characterisation such films. The impact of the ferroelectric layer on the deep defect level (3 eV below $E_{c, \min, HfO_2}$, as seen in Figure 7D) has not thusfar been explored, but if any, it could be expected to play a role during the erase phase negative voltage pulsing. Such behaviour would be consistent with observations of negative charge trapping on logic devices under negative bias stress [10].

To conclude, an extensive study of defect levels in ferroelectric Si-doped HfO_2 using a combination of trapping and photoemission techniques reveals high densities close to the silicon conduction band edge, which are readily accessible during device operation. Their presence becomes more evident with the onset of polarisation of the ferroelectric layer, which can lead to their appearing at lower energy levels than the paraelectric- HfO_2 structures they are shown to closely resemble when this contribution is decoupled.

Data Availability

The data that support the findings of this study are available from the corresponding author upon reasonable request.

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References

- [1] T. S. Bösccke, J. Müller, D. Bräuhäus, U. Schröder, U. Böttger, "Ferroelectricity in Hafnium Oxide: CMOS compatible Ferroelectric Field Effect Transistors", *IEEE International Electron Devices Meeting*, (2011), pp. 24.5.1-24.5.4, <https://doi.org/10.1109/IEDM.2011.6131606>
- [2] J. Müller, T. S. Bösccke, S. Müller, E. Yurchuk, P. Polakowski, J. Paul, D. Martin, T. Schenk, K. Khullar, A. Kersch, W. Weinreich, S. Riedel, K. Seidel, A. Kumar, T. M. Arruda, S. V. Kalinin, T. Schlösser, R. Boschke, R. van Benthum,

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- U Schröder, T Mikolajick, "Ferroelectric hafnium oxide: A CMOS-compatible and highly scalable approach to future ferroelectric memories", *IEEE International Electron Devices Meeting*, (2013), pp. 10.8.1-10.8.4, <https://doi.org/10.1109/IEDM.2013.6724605>
- [3] J Müller, P Polakowski, S Mueller, T Mikolajick, "Ferroelectric hafnium oxide based materials and devices: Assessment of current status and future prospects", *ECS Journal of Solid State Science and Technology*, **4**, (2015), ppN30-N35, <https://doi.org/10.1149/2.0081505jss>
- [4] Y Higashi, N Ronchi, B Kaczer, K Banerjee, S McMitchell, B O'Sullivan, S Klima, A Minj, U Celano, L DiPiazza, M Suzuki, D Linten, J Van Houdt, "Impact of Charge Trapping on Imprint and its Recovery in HfO₂ based FeFET", *IEEE International Electron Devices Meeting*, (2019), pp. 15.6.1-15.6.4, <https://doi.org/10.1109/IEDM19573.2019.8993472>
- [5] B Kaczer, T Grasser, PJ Roussel, J Martin-Martinez, R O'Connor, BJ O'Sullivan and G Groeseneken, "Ubiquitous relaxation in BTI stressing—new evaluation and insights", *Proceedings of IEEE International Reliability Physics Symposium*, (2008), pp20-27, <https://doi.org/10.1109/RELPHY.2008.4558858>
- [6] MH Park, C-C Chung, T Schenk, C Richter, M Hoffmann, S Wirth, JL Jones, T Mikolajick, U. Schroeder, "Origin of Temperature-Dependent Ferroelectricity in Si-Doped HfO₂", *Advanced Electronic Materials*, **4**, 4, (2018), 1700489, <https://doi.org/10.1002/aelm.201700489>
- [7] T Grasser, (2014), 'The Capture/Emission Time Map Approach to the Bias Temperature Instability', in *Bias Temperature Instability for Devices and Circuits*, edited by T Grasser, Springer, pp447-481, <https://doi.org/10.1007/978-1-4614-7909-3>
- [8] V Putcha, J Franco, A Vais, S Sioncke, B Kaczer, D Linten, G Groeseneken, "On the Apparent Non-Arrhenius Temperature Dependence of Charge Trapping in III-V/High-k MOS Stack", *IEEE Transactions on Electron Devices*, **65**, 9, (2018), pp3689-3896, <https://doi.org/10.1109/TED.2018.2851189>
- [9] R Degraeve, M Cho, B Govoreanu, B Kaczer, MB Zahid, J Van Houdt, M Jurczak, G Groeseneken, "Trap Spectroscopy by Charge Injection and Sensing (TSCIS): A quantitative electrical technique for studying defects in dielectric stacks", *2008 IEEE International Electron Devices Meeting*, (2008), pp. 1-4, <https://doi.org/10.1109/IEDM.2008.4796812>
- [10] G Rzepa, J Franco, BJ O'Sullivan, A Subirats, M Simicic, G Hellings, P Weckx, M Jech, T Knobloch, M Walzl, P Roussel, D Linten, B Kaczer, T Grasser, "Comphy - A Compact-Physics Framework for Unified Modeling of BTI", *Microelectronics Reliability*, **85** (2018), pp49-65, <https://doi.org/10.1016/j.microrel.2018.04.002>
- [11] MNK Alam, B Kaczer, L-Å Ragnarsson, M Popovici, G Rzepa, N Horiguchi, M Heyns, J Van Houdt, "On the Characterization and Separation of Trapping and Ferroelectric Behavior in HfZrO FET", *Journal of the Electron Devices Society*, **7**, (2019), pp855-862, <https://doi.org/10.1109/JEDS.2019.2902953>
- [12] F Cerbu, O Madia, DV Andreev, S Fadida, M Eizenberg, L Breuil, JG Lisoni, JA Kittl, J Strand, AL Sluger, VV Afanas'ev, M Houssa, A Stesmans, "Intrinsic electron traps in atomic-layer deposited HfO₂ insulators", *Applied Physics Letters*, **108**, (2016), 222901, <https://doi.org/10.1063/1.4952718>
- [13] J Strand, M Kaviani, D Gao, A-M El-Sayed, VV Afanas'ev, AL Shluger, "Intrinsic charge trapping in amorphous oxide films: status and challenges", *Journal of Physics: Condensed Matter*, **30**, 23, (2018), <https://doi.org/10.1088/1361-648X/aac005>

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	$\langle E_t \rangle \pm \sigma_{E_t}$	$\langle S \rangle \pm \sigma_S$	N_{ot}	R		$\langle E_t \rangle \pm \sigma_{E_t}$	$\langle S \rangle \pm \sigma_S$	N_{ot}	R
	[eV]	[eV]	[cm ⁻³]			[eV]	[eV]	[cm ⁻³]	
IL	0.44±0.25	2.2±1.6	1×10 ¹⁹	0.44	IL	0.44±0.25	2.2±1.6	1×10 ¹⁹	0.44
HfO ₂	0.33±0.05	0.05±0.01	7×10 ¹⁹	0.7	Si:HfO ₂	0.33±0.1	0.05±0.01	5×10 ¹⁹	1.0

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