Defect Engineering to Achieve Wake-up Free HfO$_2$-Based Ferroelectrics

Alireza Kashir,∗ Seungyeol Oh, and Hyunsang Hwang∗

Wake-up effect is still an obstacle in the commercialization of hafnia-based ferroelectric thin films. Herein, the effect of defects, controlled by ozone dosage, on the field cycling behavior of the atomic layer deposited Hf$_{0.5}$Zr$_{0.5}$O$_2$ (HZO) films is investigated. A nearly wake-up free device is achieved after reduction of carbon contamination and oxygen defects by increasing the ozone dosage. The sample which is grown at 30 s ozone pulse duration shows about 97% of the woken-up $P_r$ at the pristine state whereas that grown below 5 s ozone pulse time shows a pinched hysteresis loop, that underwent a large wake-up effect. This behavior is attributed to the increase in oxygen vacancy and carbon concentration in the films deposited at insufficient O$_2$ dosage, which is confirmed by X-ray photoelectron spectroscopy (XPS). The X-ray diffraction (XRD) scan shows that the increase in ozone pulse time yields the reduction of tetragonal phase; therefore, the dielectric constant reduces. The $I–V$ measurements reveal the increase in current density as the ozone dosage decreases, which might be due to the generation of oxygen vacancies in the deposited film. Finally, the dynamics of wake-up effect is investigated, and it appears to be explained well by the Johnson–Mehl–Avrami–Kolmogoroff model, which is based on structural phase transformation.

1. Introduction

Ferroelectricity in HfO$_2$-based thin films was observed first time by Boscke et al.[1] and thereafter, due to its numerous advantages such as simple structure, strong binding energy between oxygen and transition metal ions, wide bandgap ($\approx$5.3–5.7 eV), and more importantly, the compatibility with current complementary metal oxide semiconductor (CMOS) technologies, extensive research has been conducted to achieve a reliable property for various potential applications such as ferroelectric memory, ferroelectric field-effect transistors (Fe FETs), energy harvesters, pyroelectric sensors, and so on.

Bulk HfO$_2$ crystallizes in the monoclinic (P2$_1$/c) structure (m-phase) at ambient condition.[2] A martensitic phase transformation from the monoclinic to tetragonal (P4$_2$/nmc) (t-phase) is observed at $\approx$1700 °C. A further increase in temperature to $\approx$2200 °C causes a diffusionless tetragonal to cubic Fm$ar{3}$m (c-phase) transition.[3] These transition temperatures can be substantially altered by doping, mechanical stress, or surface manipulation and the high-temperature phases are practically achieved in HfO$_2$ thin films at room temperature (RT).[3–14] A non-centrosymmetric polar orthorhombic phase (Pca$_2_1$, o-phase) is believed to be the structural origin of ferroelectricity in HfO$_2$-based thin films.[2,15] Pca$_2_1$ is extremely close in free energy ($< k_B T/5$, where $k_B$ is the Boltzmann constant) to the equilibrium nonpolar phases.[2] Therefore, tuning experimental conditions (e.g., via strain or dopants) may stabilize this polar polymorph. In fact, the transition of the t-phase to o-phase instead of m-phase is a response to the large volume expansion of the crystal by $\approx$5% during later transition under the capping layer confinement. The t- to o phase transition causes $\approx$1% decrease in volume.[8] Batra et al.[3] predicted a transition from the nonpolar to polar Pca$_2_1$ phase by applying a strong electric field combined with the application of an appropriate strain. Therefore, the electric field can provide an additional driving force to induce ferroelectricity in HfO$_2$-based thin films.

Despite many researches on HfO$_2$-based ferroelectric thin films and obtaining strong ferroelectric properties, within the lifetime of the device, two critical problems i.e., wake-up and fatigue, can be identified, which cause serious obstacles for device operation. It was experimentally observed that the wake-up effect is due to the migration of point defects (mainly oxygen vacancies) and a transition from the nonferroelectric (m- and t-phase) to ferroelectric phase (especially at the interfacial region). Both mechanisms can be driven under the application of a strong cycling electric field. A locally distributed inhomogeneous internal field that originated from the spatially unevenly distributed charged defects, such as oxygen vacancies, can be the origin of the internal field in the pristine material.[16–21] During the electric field cycling process, the oxygen vacancies may diffuse into the bulk regions of the ferroelectric HfO$_2$-based films, in which case the wake-up process will occur. Starschich et al.[21] showed that even a DC pulse with an adequate length wakes up the doped HfO$_2$ capacitors. Thus, the total pulse length may be critical for the wake-up effect, implying a time-dependent phenomenon.

Dr. A. Kashir, S. Oh, Prof. H. Hwang
Center for Single Atom-based Semiconductor Device and Department of Materials Science and Engineering
Pohang University of Science and Technology (POSTECH)
Pohang, Republic of Korea
E-mail: kashir@postech.ac.kr; hwanghs@postech.ac.kr

The ORCID identification number(s) for the author(s) of this article can be found under https://doi.org/10.1002/adem.202000791.

DOI: 10.1002/adem.202000791
Another mechanism which is responsible for wake-up effect is the structural transition from nonferroelectric (m or t) to ferroelectric (o) phase driven by a strong electric field.\cite{22,23}

Therefore, the wake-up effect can be substantially reduced by the reduction of m- and t-phase during thin-film processing.\cite{24}

Goh et al.\cite{25} evaluated the effect of metal oxide (as an electrode) on the wake-up behavior of the Hf$0.5$Zr$0.5$O$_2$ (HZO) film and they observed an improve in wake-up behavior which could be due to the suppression of oxygen vacancy at the interfacial region. Oxide metal electrode can provide additional oxygen to the HZO layer to hinder the oxygen vacancy generation during the application of the electric field. Therefore, the reduction of oxygen vacancies (in both bulk and interfacial regions) and suppression of nonferroelectric (m and t) phases during HfO$_2$-based thin-film processing are the promising paths toward wake-up-free devices.

The ozone O$_3$ pulses during the atomic layer deposition (ALD) of HfO$_2$-based thin films have two major roles.\cite{26} The primary role is to infuse oxygen into Hf and Zr layers to form HfO$_2$ and ZrO$_2$ compounds. Thus, the pulse length is a crucial parameter to provide enough oxygen for stoichiometric phase formation and avoiding the generation of oxygen vacancies. But it should be noticed that the C–O bond removal out of the sample by applying O$_3$ pulses can remarkably change the subsequent annealing behavior of the as-grown film.\cite{26} Carbon is an inevitable part of the ALD-grown films as the precursors are usually organic compounds. Cho et al.\cite{26} observed the appearance of the tetragonal phase as the ozone dosage decreases which was attributed to the effect of carbonate bonds on the annealing behavior of HfO$_2$ films. The C–O bonds remain between pure HfO$_2$ domains due to the incomplete chemical reaction of the precursors, which subsequently prevent the agglomeration of the nanoscale domains, which results in the stabilization of tetragonal phase.\cite{26–28} Therefore, ozone dosage can drive the wake-up effect directly by oxygen vacancy generation and indirectly by carbon-induced t-phase formation.

As a proper capping electrode, tungsten (W) shows the lowest thermal expansion coefficient (TEC) $\alpha$ among the widely used metal electrodes, inducing an in-plane tensile strain to HZO film during rapid thermal annealing (RTA), which facilitates the formation of the o-phase.\cite{29} In-plane tensile strain stresses on the c-axis of t-phase in favor of the phase change from t- to o-phase. Moreover, WO$_3$ and WO$_2$ formation enthalpy is almost two times higher than that of HfO$_2$ and ZrO$_2$, preventing the formation of interfacial oxide during the ALD of HZO, which in turn facilitates the deposition of stoichiometric Hf$0.5$Zr$0.5$O$_2$ by an appropriate tuning of ozone dosage.

Considering these facts, we investigate the effect of O$_3$ pulse time (ozone dosage) during the ALD of the $\approx$10 nm HZO films on the wake-up behavior of W/HZO/W capacitors. The both major roles of oxygen will be considered. We will precisely study the effect of the remaining carbon and O$_3$ pulse length on the ferroelectric behavior of the HZO thin films. As we discussed, the kinetics of wake-up effect is explained by the Johnson–Mehl–Avrami–Kolmogoroff model.

2. Results and Discussion

Figure 1 shows the pristine $P$–$E$ hysteresis curves for the samples grown at different $t_{O_3}$. Increasing $t_{O_3}$ results in a more open (depinched) loop. Therefore, the pristine $2P_0$ value increases as the samples are grown at longer ozone pulse lengths. The $P$–$E$ curves at the wake-up state for two samples deposited at $t_{O_3} = 5$ and 30 s are shown in Figure 2. It revealed that the sample which was grown at lower $t_{O_3}$ undergoes a large wake-up effect, whereas the sample grown at higher $t_{O_3}$ is almost wake-up free and its pristine $2P_0$ value is about 97% of the wake-up state (Table 1). This is only 63% in case of the sample grown at $t_{O_3} = 5$ s. Lee et al.\cite{19} using Monte Carlo simulation demonstrated that the ferroelectric hysteresis of a doped HfO$_2$ thin film with a lower defect ratio exhibits an increase in both $P_r$ and the coercive field $E_c$. Defects can pin the ferroelectric domains and strongly affect the ferroelectric properties of thin films. As we discussed,

![Figure 1. Polarization–Electric field curves of $\approx$10 nm HZO films grown at different ozone pulse durations.](image)

![Figure 2. The pristine ($P$) and woken-up ($W$) Polarization–Electric field curves of $\approx$10 nm HZO films grown at O$_3$ pulses.](image)

| Ozone pulse duration | Pristine to woken-up $2P_0$ |
|----------------------|-----------------------------|
| 2 s                  | 63%                         |
| 5 s                  | 63%                         |
| 15 s                 | 90%                         |
| 30 s                 | 97%                         |

Table 1. The pristine to woken-up $2P_0$ ratio of the HZO films grown at different ozone dosages.
two major sources which led to the wake-up effect were recognized, the charged oxygen vacancies and the structural phase transition from nonferroelectric to ferroelectric phase. On the other hand, the oxygen vacancy movement during field cycling causes a local phase transition in different regions, as the stability of different polymorphs in HfO₂-based thin films depends on the amount of oxygen vacancies.\cite{10,31} In a recent work, based on the density fractional theory calculations, it was predicted that even m-phase, which is originally nonferroelectric phase, shows ferroelectricity by the introduction of oxygen vacancy.\cite{32} Therefore, the oxygen vacancy and its movement during electric field cycling contribute majorly in the wake-up effect (either by pinning of ferroelectric domain walls or by stabilizing o-phase). The decrease in O₃ dosage during the deposition of HZO might induce point defects, dominantly oxygen vacancies, and subsequently trigger the mechanisms of wake-up during electric field cycling.

**Figure 3** shows the effect of O₃ dosage on the leakage current of HZO samples. At 1 MV cm⁻¹, the sample which was grown at 5 s O₃ pulse shows an electric current density ≈ 2.5 times higher than that of the sample grown at 30 s. This behavior implies the role of ozone dosage in the reduction of defects and subsequently improves the insulating characteristics of HZO films.

It should be noted that the removal of C—O bonds out of the sample by applying O₃ pulse can remarkably influence the subsequent annealing behavior of the as-grown film.\cite{26–28} This behavior becomes much more important and crucial when the ferroelectricity of HfO₂-based thin film is the major concern. **Figure 4** shows the XPS spectra of the samples grown at two O₃ pulse times.

**Figure 3.** The current density versus electric field of HZO films grown at different O₃ pulse times.

**Figure 4.** a) Zr 3d, b) Hf 4f, c) O 1s, and d) C 1s XPS spectra of the as-grown HZO thin films deposited at different O₃ pulse times.
different O₃ pulse durations. The Zr/Hf ratio was almost 1 for both samples, indicating the (Zr₀.₅Hf₀.₅)O₂₋ₓ composition. In case of the sample grown at t₀₃ = 5 s, we detected a smaller fraction of oxygen (≈1%), compared with the sample which was grown at t₀₃ = 30 s (Figure 4c). The sample with more oxygen deficiency shows a higher amount of carbon (Figure 4d). The XPS measurement revealed around 2.1% carbon in the sample, which was grown at t₀₃ = 5 s, whereas the sample grown at a higher O₃ dosage was almost free of carbon. These measurements reveal the role of O₃ pulse time in the generation of oxygen defects and carbon contaminants in the atomic layer deposited films.

It was shown that the less O₃ pulse time causes a higher amount of carbon in the as-grown sample because of the insufficient precursor oxidation, which subsequently influences the grain size of the crystallized film. The smaller grain size would result in the t-phase formation due to its lower grain boundary energy compared with m- and even o-phases (Figure 5).[26–28] The formation of t-phase and its transformation to o-phase under strong electric field is another cause for the wake-up effect. Therefore, applying longer O₃ pulses can decrease the amount of oxygen defects and carbon content in the as-grown film, which consequently hinder the mechanisms of the wake-up effect.

XRD scans clearly demonstrate the decrease of t-phase and increase of o-phase with increasing the ozone dosage during growth (Figure 6). For clear visualization, we showed a limited range of 2θ, in which the evolution of HZO o- and t-phase Bragg peaks versus ozone dosage is clearly observable. Increasing the ozone dosage from 5 to 30 s results in a decrease in t-phase from 29% to 7%, which considerably affects the ferroelectric properties (Table 2). On the other hand, the existence of t-phase introduces more grain boundaries, which are accompanied by more defective structures. These defects can cause the wake-up effect by pinning the ferroelectric domain walls.[24]

![Figure 5. Schematic of the effect of carbon contamination (※) on the stabilization of the t-phase in a HZO thin film. Carbon removal by sufficient oxygen supply results in the optimum growth of grain to form the ferroelectric o-phase during the cooling step of the annealing process.](image)

![Figure 6. X-ray diffraction patterns of the HZO films grown at different O₃ pulse time: a) 5 s, b) 15 s, c) 30 s.](image)
The formation of the m-phase was totally suppressed during RTA under the capping electrode confinement (not shown here). All films contained t- and o-phases, possibly arising from the in-plane tensile strain applied to the HZO film during thermal annealing. Mechanical stress induced by the W-capping layer during RTA is believed to be the driving force for the t- to o-phase transition.[29] In fact, the in-plane tensile strain, which is developed during the annealing process, substantially suppresses the formation of the m-phase by preventing the necessary twin deformations required to form the m-phase. Therefore, the TEC of the capping material substantially alters the structural properties of HZO films.

The capacitance versus voltage measurement shows that the sample which was grown at a lower O₃ pulse time has a higher dielectric constant \( \varepsilon_r \), which can be partially due to the space-charge polarization induced by defects (Figure 7a). The initial increase in permittivity with DC field is probably due to the increased movement of domain walls, which become free from defects which lock them at the zero-DC field. From these points of view, the sample grown at lower ozone dosages contains more defects. Increasing \( t_{O_3} \) during growth results in HZO films with a lower dielectric constant \( \varepsilon_r \). This is in agreement with the XRD result. It was already observed that carbon stabilizes the tetragonal phase in the HfO₂-based thin film and subsequently increases \( \varepsilon_r \).[10–12] The t-phase has a higher \( \varepsilon_r \) than m- and even o-phase in HZO films. This behavior combined with the \( P-E \) measurement (Figure 7b) suggests two possible interpretations. The samples which were grown at higher oxygen dosages contain less oxygen vacancy, therefore, less density of built-in bias electric field and finally undergo less wake-up effect.

The another possible explanation is the carbon-induced t-phase formation which can consequently cause the phase transition to o-phase during field cycling. On the other hand, as mentioned previously, the oxygen vacancy movement provides a driving force for phase transition during field cycling from t- to o-phase. Finally, the effect of oxygen vacancy-induced ferroelectricity in m-phase should be considered. Therefore, O₃ pulse duration shows a critical contribution to the deposition of wake-up free devices through impeding different possible mechanisms of wake-up effects.

To investigate the nature of the wake-up effect, we studied the dynamics of wake-up behavior in different samples. Figure 8a shows the evolution of \( P_r \) of a sample, which was grown at the lowest O₃ pulse duration (2 s), under electric field cycles. The number of cycles to reach a given \( P_r \) strongly depends on the applied frequency. As the frequency of the applied electric field increases, the number of cycles for achieving a given \( P_r \) increases. By plotting the \( P_r \) over the duration of the applied field, i.e., \( \frac{\text{cycle dependence}}{\text{frequency}} \), the curves at different frequencies show the same dependence. Thus, not the amount of cycles but the duration of the applied electrical field is essential for the wake-up. That means that 100 cycles at 100 Hz show the same \( P_r \) as 500 cycles at 500 Hz. Figure 8b shows that the evolution of \( P_r \) versus \( \frac{\text{cycle dependence}}{\text{frequency}} \) is totally independent of the applied frequency and it is a time-dependent phenomenon.[21] It can be the direct electric field-induced structural change, electric field-driven oxygen vacancy movement (redistribution), or oxygen vacancy-induced phase transition. All these three mechanisms are driven by the cycling field and the progress of these processes is time dependent. The increase in coercive field \( E_c \) for different samples also shows the same behavior. Here, we plot the \( E_c \) versus \( t_{app} \) for the sample, which was grown at 30 s O₃ pulse time as an example (Figure 8c). Considering the structural change during electric field cycling (either through oxygen redistribution or direct electric field-driven structural transformation) as the major reason for wake-up effect, we investigate the kinetics of the phase transition explained by the Johnson–Mehl–Avrami–Kolmogoroff (JMAK) model[14] (Equation (1)).

---

**Table 2.** The fraction of tetragonal phase in HZO film deposited at different ozone dosages.

| Ozone pulse duration | 5 s  | 15 s | 30 s |
|----------------------|-----|-----|-----|
| Tetragonal phase fraction | 29% | 16% | 7% |

---

**Figure 7.**

(a) The pristine dielectric constant versus electric field of the HZO thin film grown at different O₃ pulse times measured at 10 kHz.

(b) The changes in pristine remnant polarization \( 2P_r \) (blue) (calculated from Figure 1) and dielectric constant \( \varepsilon_r \) (red) of HZO films versus O₃ pulse time.
where \( f \) is the fraction of transferred volume, \( t \) is time, \( k \) is a constant, and \( n \) is an integer or half-integer which is called Avrami exponent and describes the rate and geometry of nucleation and growth. Therefore

\[
\ln(-\ln(1-f)) = \ln(k) + n \ln(t)
\]  

We correlated the increase in \( P_r \) to \( f \) using Equation (3)

\[
f = \frac{P_r - P_{pri}}{P_W - P_{pri}}
\]  

where \( P_{pri} \) and \( P_W \) are the values of \( P_r \) at pristine and wake-up states, respectively. Therefore

\[
f = \begin{cases} 
0, & \text{at pristine state} \\
1, & \text{at wake-up state}
\end{cases}
\]

Figure 9 shows a linear plot for different samples during wake-up transition with almost the same slope (≈0.5). This behavior indicates that phase transition during field cycling might have a dominant role in wake-up effect. The phase transition during
the wake-up process can originate from different mechanisms. A strong electric field induces structural change in hafnia-based thin films. Indeed, the transition from m- and t- to o-phase was observed by different research groups,[12,23] which increases \( P_r \) in HfO\(_2\)-based ferroelectrics. Moreover, the oxygen vacancy diffusion into different regions changes the relative stability of polymorphs and consequently causes phase transition. It can stabilize the ferroelectric phase and consequently increase the remnant polarization and wake up the ferroelectric HZO.

3. Conclusion

The ferroelectric behavior of the \( \approx 10 \text{ nm} \) ALD-grown HZO thin films at different ozone pulse durations was investigated. To induce an in-plane tensile stress and achieve m-phase-free thin films, we used W electrode, which has a relatively low thermal expansion coefficient. It was observed that the samples which were deposited at \( t_0 = 5 \text{ s} \) underwent a large wake-up effect. The XRD pattern of all samples revealed the total suppression of m-phase formation during RTA. The fraction of t-phase decreased as \( t_0 \) increased which subsequently reduced \( \varepsilon_r \) of HZO films. The XPS study revealed \( \approx 2.1\% \) carbon in the sample grown at \( 5 \text{ s} \) O\(_3\) pulse time whereas the sample which was grown at \( 30 \text{ s} \) O\(_3\) pulse time was almost free of carbon. The carbon can stabilize the tetragonal phase. The leakage current was reduced by increasing \( t_0 \) which could be due to the decrease of defects in the HZO film. Therefore, both carbon contaminants and oxygen vacancies caused a large wake-up effect in the samples grown in an insufficient ozone environment. Tuning the oxygen pulse length during the deposition of HZO, an almost wake-up-free ferroelectric thin film was achieved. Finally, we investigated the kinetics of the wake-up effect based on the phase transition explained by the JMak model. All samples followed the JMak model with almost the same slope. This behavior suggests phase transition as a major cause of wake-up effect.

4. Experimental Section

The \( \approx 10 \text{ nm} \) HZO films were deposited on the 50 nm-thick W bottom electrode sputtered on the SiO\(_2\)/Si substrate using the ALD technique. Different ozone pulse durations \( t_0 = 2–30 \text{ s} \) were applied for each deposition. The substrate temperature was kept at 250°C during the deposition of all films in this work. The \( \text{Hf}[\text{N}-(\text{C}_2\text{H}_3\text{CH}_2)]_4 \) and \( \text{Zr}[\text{N}-(\text{C}_2\text{H}_3\text{CH}_2)]_4 \) precursors were used as the Hf and Zr metals sources, respectively. The growth rate of HfO\(_2\) and ZrO\(_2\) were almost same \( (\approx 1 \text{ Å} \text{cycle}^{-1}) \). After deposition, all films were capped with 50 nm W electrode using the radio frequency sputtering technique to fabricate metal-insulator-metal (MIM) capacitors. Finally, W/HZO/W capacitors with different electrode areas were passed through an annealing process in N\(_2\) ambient at 500°C for 30 s.

The crystalline structures of films were investigated using an X-ray diffractometer (XRD) within grazing incidence geometry. For elemental analysis, we used X-ray photoelectron spectroscopy (XPS) on as-grown samples (not capped). The ferroelectric properties of MIM capacitors were measured by the LCII ferroelectric precision tester (Radiant Technologies) and the dielectric permittivity and \( \varepsilon'-\varepsilon'' \) characteristics were evaluated using Keysight B1500A semiconductor device parameter analyzer. All measurements were carried out at RT.

Acknowledgements

This work was supported by the National Research Foundation of Korea funded by the Korea government (MSIT), grant no. NRF-2018R1A3B1052693.

Conflict of Interest

The authors declare no conflict of interest.

Keywords

carbon contents, defect engineering, ferroelectrics, HfO\(_2\), Johnson–Mehl–Avrami–Kolmogoroff model, leakage currents, wake-up effects, X-ray photoelectron spectroscopies

Received: July 3, 2020
Revised: August 4, 2020
Published online: September 28, 2020

[1] T. S. Böscke, J. Müller, D. Bräuhaus, U. Schröder, U. Böttger, Appl. Phys. Lett. 2011, 99, 102903.
[2] T. D. Huan, V. Sharma, G. A. Rossetti, R. Ramprasad, Phys. Rev. B - Condens. Matter Mater. Phys. 2014, 90, 064111.
[3] M. H. Park, Y. H. Lee, H. J. Kim, Y. J. Kim, T. Moon, K. Do Kim, J. Müller, A. Kersch, U. Schroeder, T. Mikolajik, C. S. Hwang, Adv. Mater. 2015, 27, 1811.
[4] A. Parija, G. R. Waetzig, J. L. Andrews, S. Banerjee, J. Phys. Chem. C 2018, 122, 25709.
[5] R. Batra, T. D. Huan, J. L. Jones, G. Rossetti, R. Ramprasad, J. Phys. Chem. C 2017, 121, 4139.
[6] M. H. Park, Y. H. Lee, T. Mikolajik, U. Schroeder, C. S. Hwang, Adv. Electron. Mater. 2019, 5, 1800522.
[7] M. Shandalov, P. C. McIntyre, J. Appl. Phys. 2009, 106, 084322.
[8] M. Dogan, N. Gong, T. P. Ma, S. Ismail-Beigi, Phys. Chem. Chem. Phys. 2019, 21, 12150.
[9] R. Batra, H. D. Tran, R. Ramprasad, Appl. Phys. Lett. 2016, 108, 172902.
[10] M. H. Park, Y. H. Lee, H. J. Kim, Y. J. Kim, T. Moon, K. Do Kim, S. D. Hyun, T. Mikolajik, U. Schroeder, C. S. Hwang, Nanoscale 2018, 10, 716.
[11] R. Materlik, C. Kuneth, A. Kersch, J. Appl. Phys. 2015, 117, 134109.
[12] A. Navrotsky, J. Mater. Chem. 2005, 15, 1833.
[13] P. Fan, Y. K. Zhang, Q. Yang, J. Jiang, L. M. Jiang, M. Liao, Y. C. Zhou, J. Phys. Chem. C 2019, 123, 21743.
[14] G. Sharma, S. V. Ushakov, A. Navrotsky, J. Am. Ceram. Soc. 2018, 101, 31.
[15] X. Sang, E. D. Grimley, T. Schenk, U. Schroeder, J. M. LeBeau, Appl. Phys. Lett. 2015, 106, 162905.
[16] T. Schenk, U. Schroeder, M. Pešić, M. Popovici, Y. V. Pershin, T. Mikolajik, ACS Appl. Mater. Interfaces 2016, 8, 19744.
[17] T. Schenk, M. Hoffmann, J. Ocker, M. Pešić, T. Mikolajik, U. Schroeder, ACS Appl. Mater. Interfaces 2015, 7, 20224.
[18] M. Pešić, F. P. C. Fengler, L. Larcher, A. Padovani, T. Schenk, E. D. Grimley, X. Sang, J. M. LeBeau, S. Slesazeck, U. Schroeder, T. Mikolajik, Adv. Funct. Mater. 2016, 26, 4601.
[19] T. Y. Lee, K. Lee, H. H. Lim, M. S. Song, S. M. Yang, H. K. Yoo, D. I. Suh, Z. Zhu, A. Yoon, M. R. Macdonald, X. Lei, H. Y. Jeong, D. Lee, K. Park, J. Park, S. C. Chae, ACS Appl. Mater. Interfaces 2019, 11, 3142.
[20] M. Pešić, S. Slesazeck, T. Schenk, U. Schroeder, T. Mikolajick, Phys. Status Solidi Appl. Mater. Sci. 2016, 213, 270.
[21] S. Starschich, S. Menzel, U. Böttger, Appl. Phys. Lett. 2016, 108, 032903.
[22] E. D. Grimley, T. Schenk, X. Sang, M. Pešić, U. Schroeder, T. Mikolajick, J. M. LeBeau, Adv. Electron. Mater. 2016, 2, 1600173.
[23] A. Chouprik, S. Zakharchenko, M. Spiridonov, S. Zarubin, A. Chernikova, R. Kirtaev, P. Buragohain, A. Gruverman, A. Zenkevich, D. Negrov, ACS Appl. Mater. Interfaces 2018, 10, 8818.
[24] J. Bouaziz, P. R. Romeo, N. Baboux, and B. Vilquin, ACS Appl. Electron. Mater. 2019, 1, 1740.
[25] Y. Goh, S. H. Cho, S. H. K. Park, S. Jeon, Nanoscale 2020, 12, 9024.
[26] D. Y. Cho, H. S. Jung, I. H. Yu, J. H. Yoon, H. K. Kim, S. Y. Lee, S. H. Jeon, S. Han, J. H. Kim, T. J. Park, B. G. Park, C. S. Hwang, Chem. Mater. 2012, 24, 3534.
[27] B. S. Kim, S. D. Hyun, T. Moon, K. Do Kim, Y. H. Lee, H. W. Park, Y. Bin Lee, J. Roh, B. Y. Kim, H. H. Kim, M. H. Park, C. S. Hwang, Nanoscale Res. Lett. 2020, 15, 72.
[28] H.-S. Jung, S. H. Jeon, H. K. Kim, I.-H. Yu, S. Y. Lee, J. Lee, Y. J. Chung, D.-Y. Cho, N.-I. Lee, T. J. Park, J.-H. Choi, S. Han, C. S. Hwang, ECS J. Solid State Sci. Technol. 2012, 1, N33.
[29] R. Cao, Y. Wang, S. Zhao, Y. Yang, X. Zhao, W. Wang, X. Zhang, H. Lv, Q. Liu, M. Liu, IEEE Electron Device Lett. 2018, 39, 1207.
[30] M. Hoffmann, U. Schroeder, T. Schenk, T. Shimizu, H. Funakubo, O. Sakata, D. Pohl, M. Drescher, C. Adelmann, R. Materlik, A. Kersch, T. Mikolajick, J. Appl. Phys. 2015, 118, 072006.
[31] A. Pal, V. K. Narasimhan, S. Weeks, K. Littau, D. Pramanik, T. Chiang, Appl. Phys. Lett. 2017, 110, 022903.
[32] C. Liu, F. Liu, Q. Luo, P. Huang, H. X. Xu, H. B. Lv, Y. D. Zhao, X. Y. Liu, J. F. Kang, IEEE International Electron Devices Meeting, IEEE, San Francisco, CA 2018, pp. 16.4.1–16.4.4.
[33] D. Damjanovic, Rep. Prog. Phys. 1998, 61, 1267.
[34] M. C. Weinberg, D. P. Birnie III, V. A. Shneidman, J. Non-Cryst. Solids 1997, 219, 89.