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High-performance zinc tin oxide thin-film transistors via hydrogen assisted metal capping structures



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With the rapid development of extended reality such as augmented reality/virtual reality displays, the demand for high-mobility and low-temperature-processable thin-film transistors (TFTs) has grown significantly. Oxide semiconductors are promising candidates due to their excellent electrical and optical properties. Among them, zinc tin oxide (ZTO), an indium-free oxide semiconductor, offers advantages in material abundance and electrical performance. In this study, we demonstrate a low-temperature metal-induced crystallization (MIC) approach using aluminum (Al) capped ZTO films and investigate the role of hydrogen-assisted annealing in enhancing carrier mobility. The resulting ZTO TFTs achieved a field-effect mobility of $57.5 \text{ cm}^2 \text{ V}^{-1} \text{ s}^{-1}$ after thermal annealing and $107.3 \text{ cm}^2 \text{ V}^{-1} \text{ s}^{-1}$ with additional hydrogen annealing, both without a significant threshold voltage shift. Increasing the Al capping length promotes more extensive front-channel crystallization and enhances carrier mobility. TCAD simulations confirmed the formation of an additional high mobility current path in the crystallized front-channel region. These findings highlight the potential of combining MIC and hydrogen annealing for high-performance, thermally compatible oxide electronics.

The advent of amorphous indium gallium zinc oxide (*a*-IGZO) thin-film transistors (TFTs) has enabled the replacement of conventional silicon-based TFTs in active-matrix backplanes for high-resolution flat-panel displays¹. This transition is driven by the superior uniformity, high optical transparency, and compatibility with low-temperature processing of IGZO, along with its relatively high field-effect mobility ($>10 \text{ cm}^2 \text{ V}^{-1} \text{ s}^{-1}$)^{2–4}. Although IGZO TFTs exhibit lower carrier mobility than low-temperature polycrystalline silicon (LTPS), they remain highly attractive for next-generation display technologies. LTPS TFTs, with carrier mobility typically above $70 \text{ cm}^2 \text{ V}^{-1} \text{ s}^{-1}$, have been the standard for high-refresh-rate⁵. Nevertheless, their requirement for costly and energy-intensive processes such as excimer laser annealing presents obstacles to achieving large-area uniformity and reducing power consumption. In addition, the relatively high leakage current of LTPS devices ($>1 \text{ pA}$) has prompted the development of low-temperature polycrystalline oxide (LTPO) technology, which combines the mobility advantages of LTPS with the low-power operation of oxide TFTs⁶. While the LTPO process is inherently more complex, it indicates the potential of hybrid approaches and highlights the rising significance of oxide semiconductors in advanced backplane systems. With steady improvements in both mobility and stability, oxide semiconductors

are increasingly emerging as a strong candidate for large-area, cost-effective, and high-performance display applications.

Numerous studies have focused on enhancing the carrier mobility of oxide semiconductors such as In_2O_3 ⁷, SnO_2 ⁸, indium gallium oxide⁹, zinc tin oxide (ZTO)¹⁰, indium zinc tin oxide¹¹, and IGZO¹² to satisfy the performance demands of advanced display applications. Among the various strategies employed to improve mobility, low-temperature crystallization of oxide semiconductors has emerged as one of the most effective approaches^{13–15}. The crystallization typically occurs through thermal annealing at elevated temperatures, which promotes atomic rearrangement and grain growth, thereby enhancing carrier transport. For industry-standard IGZO, crystallization generally requires annealing temperatures exceeding 700°C to induce significant structural ordering¹⁶. However, such high-temperature processes pose serious limitations, especially for flexible electronics and display backplanes fabricated on polymer substrates, due to thermal incompatibility. To overcome these thermal limitations, metal-induced crystallization (MIC) has emerged as a promising technique for enabling crystallization at substantially lower temperatures¹⁷. Although higher annealing temperatures during MIC promote larger grain growth, such thermal conditions are unfavorable for integration with flexible

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substrates, polymer-based backplanes, and other temperature-sensitive platforms. Consequently, excessive thermal processing, despite enlarging crystallites, compromises the applicability of oxide TFTs across diverse electronic applications, emphasizing the importance of effective crystallization under a constrained thermal budget. Lee et al. demonstrated that using Ta as a capping layer on IGO enabled crystallization below 250 °C, resulting in a notable enhancement in field-effect mobility from $29.3 \text{ cm}^2 \text{ V}^{-1} \text{ s}^{-1}$ to $76.9 \text{ cm}^2 \text{ V}^{-1} \text{ s}^{-1}$ ¹⁸. These findings highlight the potential of MIC as a pathway for achieving high-performance oxide semiconductors under thermal constraints. Furthermore, hydrogen incorporation has also been reported to facilitate low-temperature crystallization and improve electrical performance. Magari et al. reported that hydrogenated polycrystalline In_2O_3 ($\text{In}_2\text{O}_3\text{:H}$) thin films could be crystallized at relatively low temperatures, achieving high field-effect mobility values exceeding $100 \text{ cm}^2 \text{ V}^{-1} \text{ s}^{-1}$ ¹⁹. Here, hydrogen played a crucial role in enhancing the crystallinity of oxide semiconductors by promoting grain growth and suppressing subgap defect states, thereby improving carrier transport properties.

In this paper, ZTO was employed as the active channel material, and MIC was induced by depositing an Al capping layer onto the ZTO surface, effectively crystallizing the amorphous ZTO into a cubic spinel structure. Subsequently, the influence of an intermediate hydrogen annealing step was systematically investigated by comparing the structural and electrical properties of devices with and without hydrogen treatment. Although IGZO offers excellent electron mobility and optical transparency, its structural complexity, along with the high cost and scarcity of indium, limits its scalability²⁰. Therefore, this study employed indium-free oxide semiconductor ZTO as a viable alternative. In the same context, this study adopted Al as a catalytic layer instead of expensive metals such as Ta, owing to its excellent process compatibility, low cost, and suitability for large-area fabrication. Furthermore, the electrical characteristics were examined by varying the length of the Al capping layer^{21,22}. An increase in Al length led to an extended crystallized region, which in turn enhanced the carrier mobility. The device subjected to annealing without hydrogen treatment exhibited a maximum field-effect mobility of $57.5 \text{ cm}^2 \text{ V}^{-1} \text{ s}^{-1}$, whereas hydrogen-assisted annealing yielded a significantly higher mobility of $107.3 \text{ cm}^2 \text{ V}^{-1} \text{ s}^{-1}$, with negligible shifts in threshold voltage. To gain deeper insight into this phenomenon, technology computer-aided design (TCAD) simulations were conducted with varying Al capping lengths. The amorphous and crystallized regions were modeled as distinct layers to enable a more detailed analysis of the carrier transport mechanism. The simulation results revealed the formation of two current paths: one through the back channel and another through the front channel formed by the crystallized region. As the Al length increased, the length of the high-mobility front-channel path also extended, leading to an overall increase in carrier mobility. Additionally, by independently adjusting the carrier concentration in the simulation, it was confirmed that the mobility enhancement originated from the crystallization effect rather than from carrier concentration boosting.

Results and discussion

Thin-film properties of ZTO-OAA and ZTO-HAA

Figure 1a shows the Al-induced crystallization mechanism of ZTO at the upper interfacial region. In this schematic, two distinct post-annealing conditions are introduced: ZTO-OAA (Only Air Anneal) and ZTO-HAA (Hydrogen-Assisted Anneal). Under the OAA condition, the Al-capped ZTO film was sequentially annealed in air at 300 °C and 350 °C for 1 h each. In contrast, the HAA condition included an initial annealing step in a hydrogen atmosphere at 300 °C for 1 h prior to the subsequent air annealing at 350 °C for another 1 h. Further details of the fabrication process can be found in the Methods section. The as-deposited ZTO film exhibits an amorphous structure characterized by weakly bonded oxygen species, structural disorder in bond lengths and angles, unsaturated coordination, and dangling bonds²³. Notably, during post-annealing after Al deposition, the thermal oxidation of Al leads to a distinct alteration in the chemical state

of the underlying ZTO layer. The Gibbs free energy of formation (ΔG_f) at a temperature of 350 °C for ZnO, SnO_2 , and Al_2O_3 are approximately −280, −450 and −1500 kJ/mol, respectively^{24–26}. The substantially lower ΔG_f of Al_2O_3 compared to ZnO and SnO_2 implies that Al has a higher oxidation potential than the metal components of ZTO. The oxidation of Al preferentially utilizes weakly bonded oxygen species present in ZTO, such as interstitial oxygen, hydroxyl groups, and loosely bonded oxygen, leading to the formation of AlO_x ²⁷. Through this process, the instability of the original microstructure increases, leading to structural rearrangement at 350 °C. Consequently, crystallization is enabled at a temperature lower than the inherent crystallization point of ZTO (~700 °C)²⁸. Furthermore, the incorporation of a hydrogen ambient post-annealing during the intermediate stage of thermal treatment results in a substantially enhanced change in the material characteristics. The incorporated hydrogen can exist as interstitial species or in the form of hydroxyl groups, which further increase the instability of the film²⁹. The HAA influences crystallization through well-defined defect-chemical processes rather than through an increase in structural instability. Prior DFT and ab-initio molecular dynamics studies have shown that hydrogen in metal oxide-based networks stabilizes as either covalent M-OH species or more ionic M-H-M complexes, which exhibit distinctly different effects on structural rearrangement and electronic transport³⁰. In disordered or interfacial regions, the formation and transformation of these hydrogen-related defects promote local metal-oxygen bond rearrangement, thereby facilitating crystallization, while excessive hydrogen limits grain growth^{13,31}. Weakly bonded oxygen species, including hydroxyl-related components, are preferentially consumed or reorganized near the metal/oxide interface, lowering the energy barrier for bond rearrangement and enabling microstructural ordering at reduced temperatures. After crystallization or partial ordering is established, hydrogen primarily contributes to defect-state passivation^{13,31}. Subsequent thermal treatment in an air ambient induces hydrogen out-diffusion, accompanied by structural rearrangement and crystallization. At the same annealing temperature, the ZTO-HAA condition, which includes an additional hydrogen annealing step followed by air annealing, produces a larger crystallized region than the ZTO-OAA condition.

To further validate these structural and compositional changes, the chemical bonding states of the ZTO films were analyzed using XPS, focusing on the O 1s peaks. The O 1s spectra were deconvoluted into three components: a low binding energy peak (~530 eV) attributed to lattice oxygen in M-O-M bonds (M-O), a medium binding energy peak (~531 eV) associated with oxygen adjacent to vacancies (M-O_{vac} , where O_{vac} denotes oxygen vacancies), and a high binding energy peak (~532 eV) corresponding to oxygen in hydroxyl groups (M-OH). In the single ZTO films without Al capping (Supplementary Fig. 1), both OAA- and HAA-treated samples remain amorphous and exhibit nearly identical peak compositions: ~76% of M-O, ~19% of M-O_{vac} , and ~4–5% of M-OH. This provides a baseline reference for the amorphous state prior to Al-induced crystallization. After Al capping and subsequent annealing (Fig. 1b, c), the O 1s spectra change markedly, with an increase in M-O and a corresponding decrease in M-O_{vac} and M-OH, consistent with the formation of c-ZTO at the Al/ZTO interface. The ZTO-HAA film contains more M-O (88.2%) and fewer M-O_{vac} (7.4%) species than the OAA film (83.3% and 13.0%, respectively), indicating that hydrogen-assisted pre-treatment facilitates the formation of a more ordered crystalline structure with reduced oxygen vacancies. The slightly elevated M-OH level in the ZTO-HAA film is likely due to residual hydroxyl groups introduced during hydrogen annealing.

Figure 1d shows the GIXRD patterns of ZTO films in the as-deposited state and after OAA or HAA treatment, both performed through sequential annealing at 300 °C and 350 °C but under different atmospheres. For the ZTO without an Al capping layer, no discernible peaks other than the substrate Si peak are observed, indicating its amorphous phase. In contrast, the Al-capped ZTO film shows no significant changes in the as-deposited state but exhibits distinct peak variations after annealing. These peaks appear at 17.76° , 29.32° , 34.46° , 35.97° , 41.89° , 55.48° , and 60.55° , corresponding to the (111), (220), (311), (222), (400), (511), and (440) planes,

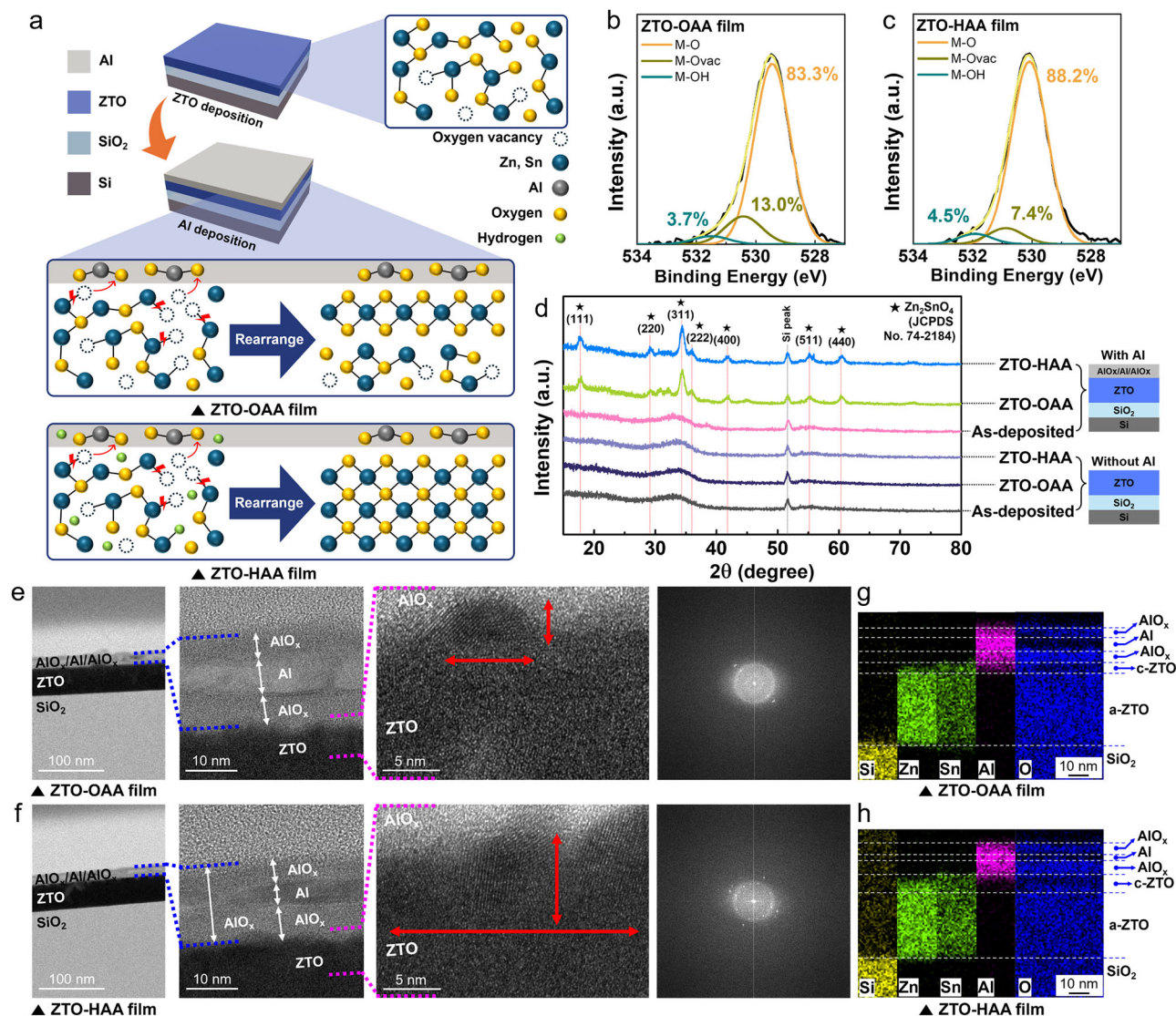


Fig. 1 | Thin-film properties. **a** Schematic of the ZTO crystallization mechanism induced by the Al layer. **b**, **c** X-ray photoelectron spectroscopy (XPS) spectra of O 1s peaks for ZTO-OAA and ZTO-HAA films, respectively. **d** Grazing incidence X-ray diffraction (GIXRD) spectra of ZTO films processed under different annealing conditions. **e**, **f** Cross-sectional high-resolution transmission electron microscopy (HR-TEM) images and corresponding Fast Fourier Transform (FFT) patterns for ZTO-OAA and ZTO-HAA films. The red arrows in (**e**, **f**) indicate the spatial extent

of the crystallized ZTO (c-ZTO) region near the Al/ZTO interface. **g**, **h** Energy-dispersive X-ray spectroscopy (EDS) mapping profiles for ZTO-OAA and ZTO-HAA films showing Si, Zn, Sn, Al, and O, respectively. All OAA and HAA treatments in this figure were carried out using a two-step annealing sequence: 300 °C for 1 h followed by 350 °C for 1 h, with OAA performed in air at both steps and HAA performed in a hydrogen ambient at 300 °C and in air at 350 °C.

respectively, and match those of Zn₂SnO₄ with a cubic spinel structure (JCPDS No. 74-2184)³². This result indicates that crystallization occurred in the films under both OAA and HAA conditions. Although the ZTO-HAA film shows slightly higher and sharper peaks than the OAA film, considering the shallow penetration depth of GIXRD at an incidence angle of 0.5°, it is difficult to clearly determine the difference in crystallized regions between the two conditions. To provide a more quantitative assessment of the crystallization properties, the crystallite size of the ZTO films was further analyzed using peak broadening analysis. To extract the crystallite size of the ZTO films, the Zn₂SnO₄ (311) GIXRD peak was analyzed using the Scherrer equation. We used the equation shown below

$$D = \frac{K\lambda}{\beta \cos \theta} \quad (1)$$

where D is the crystallite size, K is the shape factor, λ is the X-ray wavelength (Cu K α radiation), β is the full width at half maximum (FWHM) of the

diffraction peak after correction for instrumental broadening, and θ is the Bragg angle^{33–35}. Based on the corrected FWHM, the crystallite size was estimated to be ~10.3 nm for the OAA film and increased to ~11.5 nm for the HAA film. The reduced peak broadening observed in the HAA-treated sample indicates an increased coherence length, suggesting that hydrogen incorporation promotes more effective lattice ordering at the Al/ZTO interface.

Although the extracted crystallite sizes remain within the nanocrystalline regime, the substantial enhancement in electrical performance indicates that carrier transport is not dominated by grain size alone. Grain boundaries formed during crystallization act as charge trapping and scattering centers^{14,36}. However, their influence becomes secondary when improved lattice structure and reduced defect-related disorder dominate the transport behavior. In the present system, metal-induced crystallization enables the formation of a well-ordered spinel ZTO network at a relatively low temperature, while hydrogen treatment further suppresses defect states associated with grain-boundary interfaces^{13,19,37}.

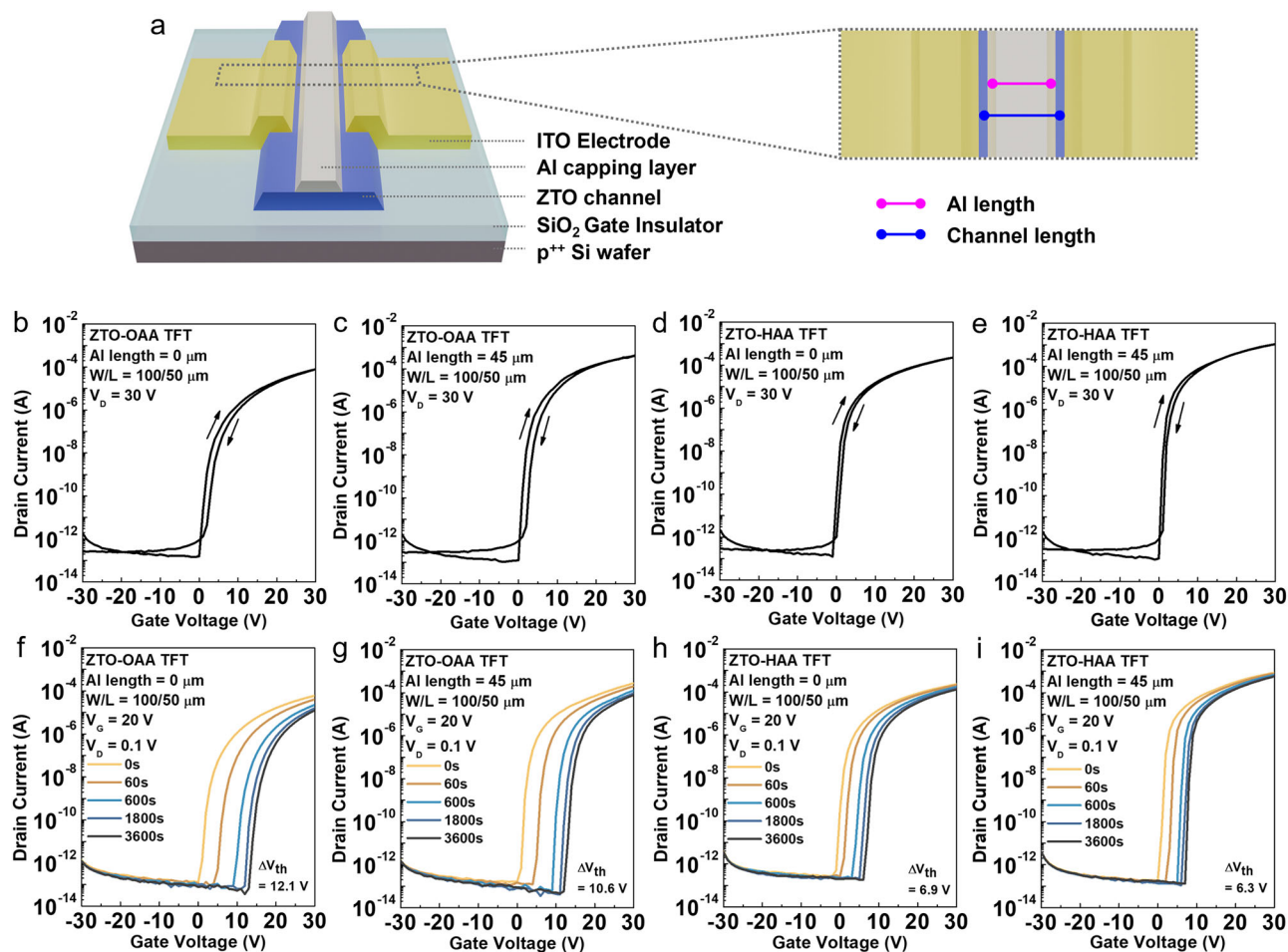


Fig. 2 | Device structure and transfer characteristics of ZTO TFTs with and without Al capping under positive bias stress (PBS). a Schematic structure of the Al-capped ZTO TFT. Transfer characteristics and their evolution under PBS for

ZTO-OAA TFTs b, f without and c, g with the Al capping layer, and for ZTO-HAA TFTs d, h without and e, i with the Al capping layer.

Cross-sectional TEM analysis was performed to directly observe the crystallized regions and complement the limited depth information obtained from GIXRD. Figure 1e, f indicate cross-sectional HR-TEM and Fast Fourier Transform (FFT) analysis of the Al-capped ZTO layer annealed under ZTO-OAA and ZTO-HAA conditions. The HR-TEM images reveal a distinct Al/ZTO/SiO₂ stacked structure. The Al layer is composed of three sub-layers: a top AlO_x layer, a central metallic Al layer, and a bottom AlO_x layer adjacent to the ZTO. The upper AlO_x layer originates from surface oxidation upon ambient exposure, while the lower AlO_x layer is formed mainly through oxidation reactions at the Al/ZTO interface during annealing, as discussed earlier. In the ZTO-HAA film, the central metallic Al layer is thinner, and regions where it is completely converted to AlO_x are also observed. High-magnification observations near the AlO_x/ZTO interface, together with the corresponding FFT patterns, show that the *c*-ZTO region in the ZTO-HAA film is more uniform, extends over a wider area, and contains a greater number of diffraction spots than in the OAA film. This suggests that, under HAA conditions, the Al undergoes more extensive interfacial oxidation with ZTO, which facilitates enhanced crystallization of the adjacent ZTO region. Furthermore, Supplementary Fig. 2 presents the HR-TEM image and corresponding FFT pattern of the ZTO region without an Al capping layer, located between the Al-capped region and the S/D electrodes, annealed under identical annealing conditions. The absence of Al in this region results in no observable diffraction spots in the FFT, indicating that the film remains amorphous. This confirms that crystallization is induced by the presence of an Al capping layer, at temperatures as low as 350 °C. The elemental distributions of Si, Zn, Sn, Al, and O in the Al-capped

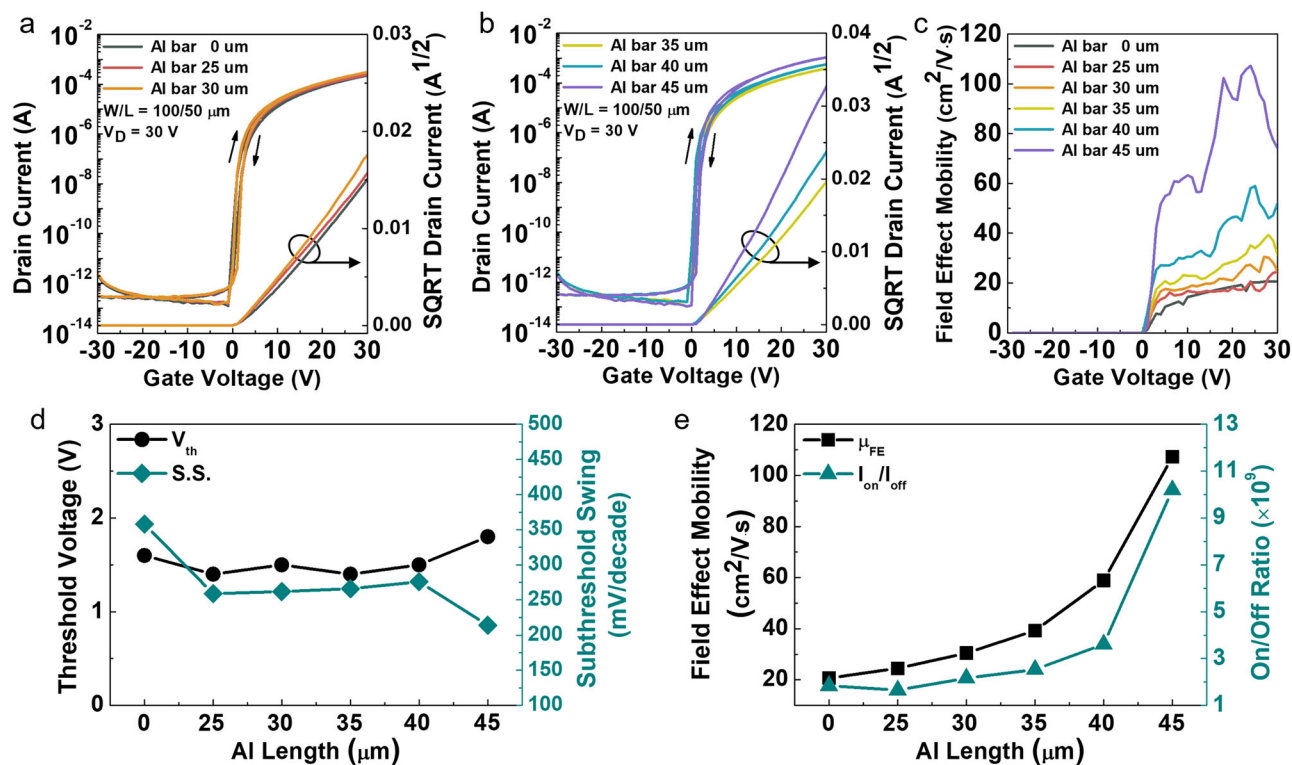
ZTO films were further analyzed by EDS mapping, as shown in Fig. 1g, h. In addition, EDS mapping data indicated that crystallized ZTO contains a lower oxygen concentration compared to amorphous ZTO, which can be attributed to the formation of oxygen vacancies induced by Al bar with a lower Gibbs free energy of oxidation. These oxygen vacancies facilitate structural reconstruction. Consistent with the HR-TEM and FFT observations, the *c*-ZTO region with lower oxygen content is thicker in the HAA-treated film. The lower oxygen content in the *c*-ZTO region compared to the amorphous region is attributed to oxygen consumption during AlO_x formation at the interface, the generation of oxygen vacancies during crystallization, and the removal of excess oxygen as the amorphous structure transforms into a stable crystalline lattice.

Electrical characteristics of ZTO-OAA and ZTO-HAA TFTs

Figure 2a illustrates the schematic structure of the Al-capped ZTO TFTs. To prevent an electrical short between the S/D electrodes, the Al capping layer was patterned as an isolated region, separated from the electrodes. The length of the Al capping layer was optimized to 45 μm for a ZTO TFT with a 50 μm channel length (Supplementary Fig. 3). Figure 2b, c show the transfer characteristics of ZTO-OAA TFTs without and with the Al capping layer, respectively. Table 1 summarizes the key performance parameters of the TFT devices, including threshold voltage (V_{th}), on/off current ratio (I_{on}/I_{off}), subthreshold swing (SS), hysteresis voltage (ΔV_{hys}), and the field-effect mobility (μ_{FE}). As a result of Al-induced crystallization, the μ_{FE} of the ZTO TFT increased from 11.5 to 57.5 cm² V⁻¹ s⁻¹ when the Al capping layer was introduced. Figure 2d, e shows the transfer characteristics of ZTO-HAA

Table 1 | Summary of the electrical parameters extracted from the gate sweep at $V_D = 30$ V of the devices

	Al length [μm]	V_{th} [V]	I_{on}/I_{off}	S.S. [mV/dec]	ΔV_{hys} [V]	μ_{FE} [$\text{cm}^2 \text{V}^{-1} \text{s}^{-1}$]
ZTO-OAA TFT	0	5.0	5.49×10^8	459	1.6	11.5
ZTO-OAA TFT	45	4.0	3.93×10^9	321	1.7	57.5
ZTO-HAA TFT	0	1.6	1.84×10^9	358	0.7	20.7
ZTO-HAA TFT	45	1.8	1.02×10^{10}	214	0.7	107.3

**Fig. 3 | Transfer characteristics and extracted electrical parameters of ZTO-HAA TFTs with different Al lengths.** **a** Transfer characteristics (I_D - V_G and $\sqrt{I_D}$ - V_G) of devices with shorter Al lengths (0–30 μm). **b** Transfer characteristics (I_D - V_G and $\sqrt{I_D}$ - V_G) of devices with longer Al lengths (35–45 μm). **c** Extracted field-effect mobility as a function of gate voltage. **d** Threshold voltage and subthreshold swing versus Al length. **e** Field-effect mobility and on/off current ratio versus Al length.

TFTs without and with the Al capping layer, respectively. In this case, the introduction of the Al capping layer markedly enhanced the mobility from 20.7 to 107.3 $\text{cm}^2 \text{V}^{-1} \text{s}^{-1}$. These results demonstrate that HAA leads to a greater improvement in mobility due to the increased crystallized regions. The incorporation of a 45 μm Al bar led to an enhancement in the I_{on}/I_{off} ratio and a decrease in SS, effects that can be attributed to oxygen vacancies generated by the Al layer, which induces structural reconstruction. Moreover, hydrogen incorporation in ZTO-HAA TFTs suppressed channel defects by filling oxygen vacancies. These results lead to improvements in both SS and hysteresis. As a result of these mechanisms, ZTO-HAA TFTs with Al capping exhibit consistently improved electrical parameters compared to ZTO-OAA TFTs.

In addition, the operational stability of the ZTO-OAA and ZTO-HAA TFTs was evaluated under positive bias stress (PBS) for 3.6×10^3 s, as shown in Fig. 2f–i. Compared to the ZTO-OAA TFTs (+12.1 and +10.6 V), the ZTO-HAA TFTs exhibit much smaller ΔV_{th} (+6.9 and +6.3 V), confirming that the HAA process effectively enhances device stability under PBS. The reason why V_{th} does not change significantly even after Al capping is that V_{th} is mainly determined by the uncapped ZTO regions adjacent to the S/D, as illustrated in Fig. 2a. Consequently, the V_{th} variation observed during stress tests is not primarily caused by Al-induced crystallization but rather by changes in interfacial and bulk trap states. Hydrogen annealing effectively reduces these trap sites by passivating weakly bonded oxygen species,

thereby lowering the density of charge trapping centers³⁸. As a result, HAA-treated devices exhibit improved PBS stability compared to OAA-treated ones, since fewer trap states are available for electron capture. This mechanism is further supported by the XPS results shown in Fig. 1c. In addition, the negative bias stress (NBS) characteristics of the ZTO-OAA and ZTO-HAA TFTs are shown in Supplementary Fig. 4. Under NBS, V_{th} variations are generally associated with oxygen vacancy migration and hole trapping, while hole trapping is intrinsically suppressed due to the low hole concentration in n-type oxide semiconductors³⁹. However, all samples exhibit relatively small negative ΔV_{th} (<1.5 V), indicating that the devices are less affected by these effects regardless of the annealing condition or Al capping. Supplementary Fig. 5 summarizes the stress time-dependent ΔV_{th} under PBS and NBS and enables a comparison of the bias stress stability. As discussed above, the variation of V_{th} under NBS remains small over stress time, whereas PBS results in a more pronounced V_{th} shift that is significantly reduced by the HAA treatment.

Further to characterize the effect of Al capping layer length on electrical performance, Al capping layers were patterned with lengths ranging from 25 to 45 μm in steps of 5 μm . All devices were fabricated under the ZTO-HAA condition. Figure 3a presents the transfer characteristics of devices with shorter Al lengths (0–30 μm). As the Al length increases, both the on-current and μ_{FE} exhibit a gradual enhancement. In Fig. 3b, devices with longer Al lengths (35–45 μm) show a more pronounced enhancement in

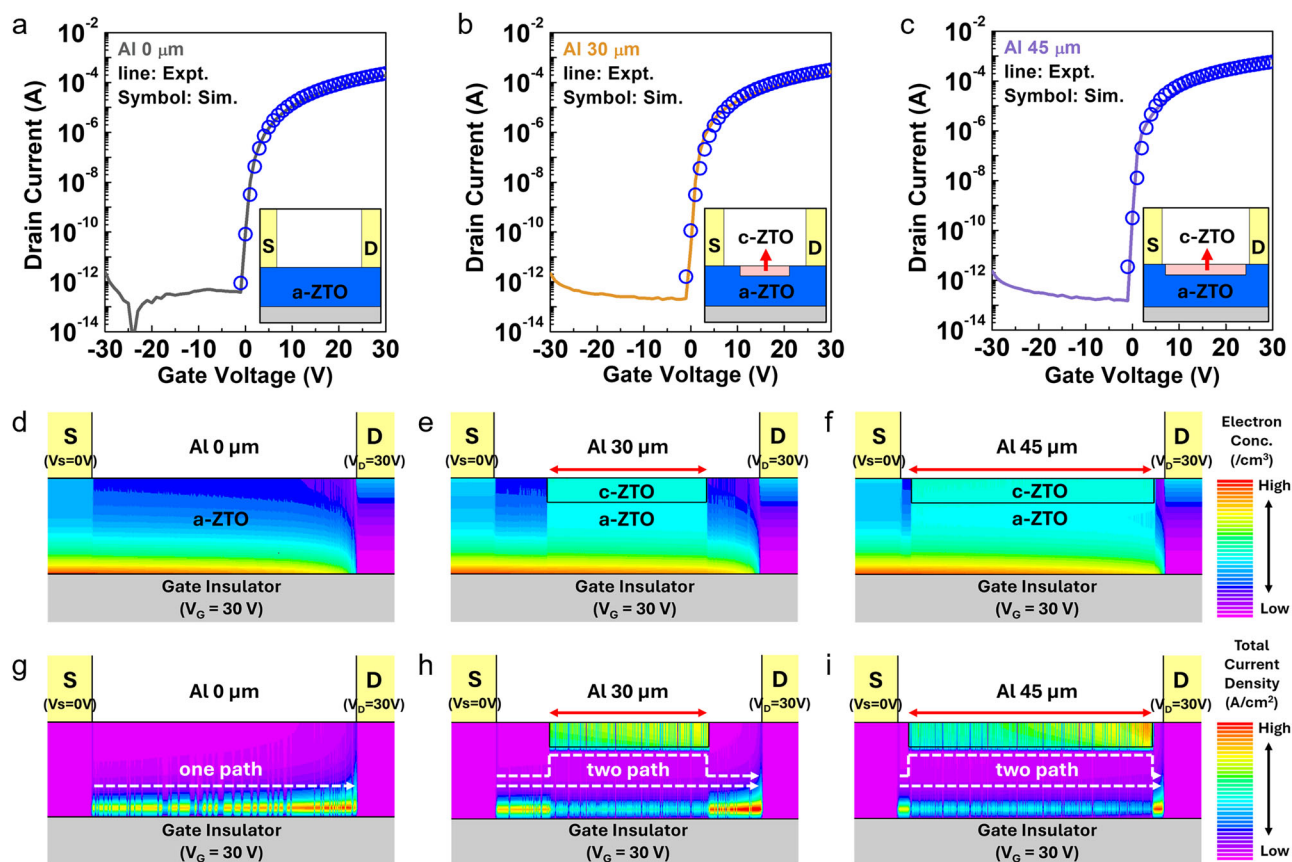


Fig. 4 | TCAD simulation of the ZTO-HAA TFTs. Experimental and simulated transfer characteristics of devices with different Al lengths: (a) 0 μm , (b) 30 μm , and (c) 45 μm . Simulated electron concentration (d–f) and total current density (g–i)

contours under identical bias conditions ($V_G = 30\text{ V}$, $V_{DS} = 30\text{ V}$) for devices with Al lengths of: (d, g) 0 μm , (e, h) 30 μm , and (f, i) 45 μm .

drain current, together with a clear separation of the $\sqrt{I_D} - V_G$ slopes. It is also noteworthy that the $\sqrt{I_D} - V_G$ plots exhibit negligible changes in V_{th} indicating that the crystallization process mainly enhances the carrier transport without significantly affecting V_{th} . Figure 3c summarizes the extracted μ_{FE} as a function of gate voltage, confirming that mobility systematically improves with increasing Al length, reaching values above $100\text{ cm}^2\text{ V}^{-1}\text{ s}^{-1}$ at 45 μm . The variations in key electrical parameters of the TFTs with different Al capping layer lengths are shown in Fig. 3d, e. According to the preceding TEM analysis, crystallization is confined to the areas directly beneath the Al capping layer on the ZTO film. Consequently, the extent of crystallization is dependent on the Al capping layer length, ranging from the smallest area at 25 μm to the largest at 45 μm . This trend is consistent with the transfer curve results, where both the on-current and μ_{FE} increase progressively with the Al capping layer length. Furthermore, the negligible change in V_{th} (less than 0.5 V) supports the occurrence of crystallization. The SS remains stable, while substantial improvements are observed in μ_{FE} (from 20.7 to $107.3\text{ cm}^2\text{ V}^{-1}\text{ s}^{-1}$) and I_{on}/I_{off} (1.84×10^9 to 1.02×10^{10}). The output characteristics and detailed device parameters with varying Al lengths are shown in Supplementary Fig. 6 and Supplementary Table 1.

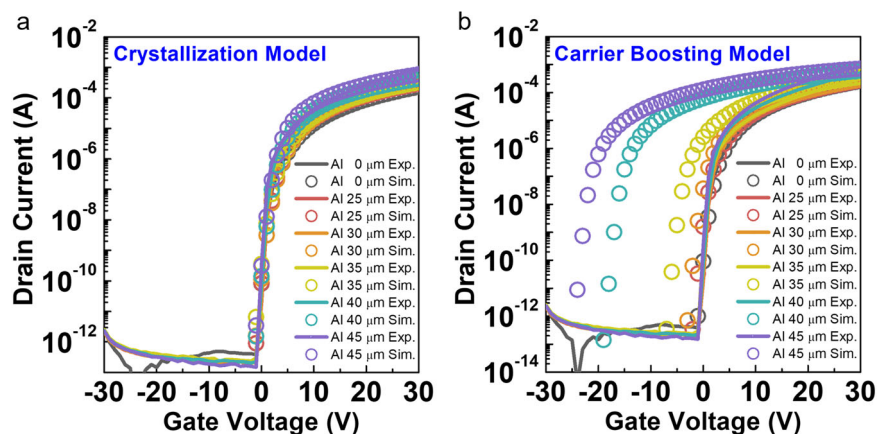
TCAD simulation of ZTO-OAA and ZTO-HAA TFTs

To elucidate the impact of Al-induced crystallization on carrier transport and device characteristics, TCAD simulations were carried out based on ZTO-HAA TFTs. The *a*-ZTO structure without an Al capping layer was modeled as a single uniform ZTO layer. In contrast, the structure with an Al capping layer, which includes both crystalline and amorphous regions, was divided into two parts: *c*-ZTO region with a length identical to that of the Al capping layer and the remaining region as *a*-ZTO. The *a*-ZTO physical

model was developed based on experimental data, while the *c*-ZTO model was constructed by modifying the *a*-ZTO parameters with a reduced density of states (DOS) and enhanced electron mobility. The extracted parameters were first obtained by fitting the *a*-ZTO-HAA TFT without Al capping (0 μm length). These parameters were then applied to the *a*-ZTO region beneath the *c*-ZTO in the *c*-ZTO-HAA TFTs with non-zero Al capping layer lengths. The detailed simulation parameters are summarized in Supplementary Table 2. Figure 4a–c and Supplementary Fig. 7 present the simulated transfer characteristics of ZTO-HAA TFTs with different Al lengths, obtained from TCAD fitting at a constant V_{DS} of 30 V. The TCAD fitting results show that as the Al length increases, the on-current rises without a noticeable shift in V_{th} , indicating a good agreement between the experimental data and simulation results. This suggests that the simulation environment effectively reproduces the behavior of the ZTO-HAA TFTs.

Based on these results, Fig. 4d–f, g–i show the simulated electron concentration and total current density contour plots, respectively, under the on-state condition ($V_G = 30\text{ V}$, $V_{DS} = 30\text{ V}$) for Al lengths of 0, 30, and 45 μm , respectively. The ZTO-HAA TFT with Al length of 0 μm is equivalent to a single-channel TFT with a bottom-gate structure, exhibiting a typical behavior where both the carrier concentration and total current density are concentrated near the front-channel region (Fig. 4d, g)⁴⁰. Interestingly, the presence of the Al capping layer leads to distinct behavior. The *c*-ZTO formed by Al-induced crystallization results in a slight elevation in electron concentration near the Al capping layer, while the back-channel *a*-ZTO region located between the Al capping layer and the S/D electrodes, where the Al capping layer is absent, still exhibits a low carrier concentration (Fig. 4e, f). In terms of total current density, which represents the current flow per unit area at specific positions within the device, the *c*-ZTO region exhibits a substantially stronger current distribution than the *a*-ZTO region

Fig. 5 | Experimental and TCAD-simulated transfer characteristics based on two models. a Crystallization model. **b** Carrier boosting model.



within the Al length. In contrast, in regions without the Al capping layer, the front-channel *a*-ZTO shows a dominant total current density distribution (Fig. h, 4i). These results demonstrate that carrier transport in Al-capped ZTO-HAA TFTs is governed by two distinct paths: the *a*-ZTO front-channel that primarily controls the V_{th} , and the *c*-ZTO back-channel induced by Al capping, which enhances the on-current. The existence of two parallel channel paths accounts for the observed increases in on-current and μ_{FE} . As the Al length increases, the length of the high-mobility *c*-ZTO path also extends, which contributes to the gradual enhancement in μ_{FE} . Based on both experimental and simulation results, the current transport mechanism of Al-induced crystallization in ZTO TFTs is clearly understood. This comprehensive understanding provides fundamental insight into how localized crystallization governs carrier pathways, offering a deeper framework for analyzing structure-property relationships in oxide semiconductors.

A further noteworthy observation is the invariance of the V_{th} , which raises important questions regarding its underlying mechanisms. The previously conducted simulation is hereafter referred to as the crystallization model. Since the primary factor for mobility enhancement is not the variation in carrier concentration, the donor concentration in the *c*-ZTO region was set equal to that of the *a*-ZTO region, thereby eliminating the influence of carrier enhancement. Therefore, as shown in Fig. 5a, both the experimental data and the transfer curves fitted using the crystallization model for various Al lengths exhibit a consistent trend of increasing on-current, while the V_{th} remains nearly unchanged. If the increase in on-current level had resulted from a sufficiently large increase in carrier concentration due to Al-induced oxygen vacancies, the interpretation would have been markedly different and it would have made it difficult to simultaneously fit the on-current level and the threshold voltage in the simulation framework. To verify this, additional TCAD simulations were conducted. Following the approach used in the crystallization model, the fitting parameters obtained from the single *a*-ZTO-HAA TFT with 0 μm Al capping layer were used as a baseline. When introducing the Al capping layer, all parameters such as the DOS and electron mobility were kept unchanged, while only the donor concentration was increased. The detailed numerical parameters are summarized in Table 2. This approach is hereafter referred to as the carrier boosting model. Figure 5b shows the transfer curves fitted using the carrier boosting model, where the fitting is performed to match the on-current level of the experimental data at $V_{DS} = 30$ V. As displayed, a clear mismatch in V_{th} between the experimental data and the simulation is observed. Notably, as the Al length increases, the V_{th} exhibits a progressively larger negative shift in simulation, but this is not experimentally observed⁴¹. By contrasting the crystallization model with the carrier boosting model, we exclude the possibility of donor-induced carrier enhancement as the primary mechanism. This comparative analysis provides compelling evidence that Al-capped ZTO-HAA TFTs achieve

Table 2 | Comparison of donor concentration ($\times 10^{17} \text{ cm}^{-3}$) extracted from transfer characteristics using two models: crystallization and carrier boosting

Al length [μm]	0	25	30	35	40	45
Crystallization model	2.5	2.5	2.5	2.5	2.5	2.5
Carrier boosting model	2.5	2.7	3.0	4.0	7.0	8.5

mobility enhancement through crystallization without V_{th} degradation, establishing a fundamental understanding of current transport in oxide TFTs.

Conclusions

In conclusion, we have reported a systematic study showing that the introduction of an Al capping layer effectively drives the crystallization of ZTO into a cubic spinel structure. Through the introduction of hydrogen annealing, crystallization is further promoted compared to devices annealed only in air at the same temperature, resulting in significant mobility enhancement. The Al capping layer length controls the extent of crystallization, leading to the formation of a dual channel path where the *a*-ZTO front-channel maintains a stable V_{th} , while the *c*-ZTO back-channel enhances the on-current. TCAD simulations validate this mechanism by distinguishing between the crystallization model and the carrier boosting model, confirming that mobility enhancement occurs without V_{th} shift. The results provide new insights into the fundamental mechanisms of MIC and suggest a promising strategy for realizing high-performance oxide TFTs with both scalability and industrial applicability.

Methods

Device fabrication

The devices were fabricated on a heavily boron-doped p-type silicon wafer, which served as the gate electrode. A 200-nm-thick thermally grown SiO_2 layer on top of the wafer acted as the gate dielectric. To remove organic contaminants, the wafer was cleaned by ultrasonic treatment in acetone and isopropyl alcohol for 10 min each. Subsequently, a 40-nm-thick amorphous ZTO layer was deposited via RF magnetron sputtering with a target ($\text{Zn}:\text{Sn} = 7:3$) at a power of 150 W under an $\text{Ar}:\text{O}_2$ flow rate of 18:2 sccm. A 15-nm-thick Al capping layer was deposited by thermal evaporation. Following Al capping, two types of annealing were applied to the ZTO channel. Under the first condition, the channel was sequentially annealed in air at 300 °C and 350 °C, each for 1 h, referred to as the ZTO-OAA (Only Air Anneal) condition. Under the second condition, the channel was annealed in a hydrogen-containing atmosphere ($\text{H}_2:\text{N}_2 = 1:9$) at 300 °C for 1 h, followed by air annealing at 350 °C for an additional 1 h, referred to as the ZTO-HAA (Hydrogen-Assisted Anneal) condition. A 40-nm-thick indium tin oxide (ITO) source/drain (S/D) electrode layer was deposited by RF

magnetron sputtering at 150 W with an Ar:O₂ flow rate of 20:0 sccm. All layers were patterned using a lift-off process. The devices were designed with a bottom-gate, top-contact configuration, incorporating a channel with a width of 100 μm and a length of 50 μm, corresponding to a width-to-length ratio (W/L) of 2. To ensure electrical isolation, the Al capping layer was designed with lengths ranging from 25 to 45 μm in 5 μm steps, introducing a defined spacing from the S/D electrodes.

Film and device characterizations

Film crystallinity was analyzed via grazing incidence X-ray diffraction (GIXRD, SmartLab, Rigaku) with a Cu Kα X-ray source. Surface and cross-sectional microstructures were investigated using optical microscopy (OM, BX53M, Olympus) and high-resolution transmission electron microscopy (HR-TEM, JEM-2100F, JEOL), respectively. The thickness of the films was measured with atomic force microscopy (AFM, NX10, Park Systems). Chemical bonding states were examined by X-ray photoelectron spectroscopy (XPS, Nexsa, ThermoFisher Scientific), while elemental composition profiles were obtained by energy-dispersive X-ray spectroscopy (EDS) during TEM analysis.

Analysis method of the TFT characteristics

The electrical characteristics of the fabricated transistors were measured using an Agilent 4156C semiconductor parameter analyzer in a dark box under ambient conditions. The μ_{FE} of the TFTs was calculated according to:

$$I_{DS} = \frac{1}{2} \frac{W}{L} \mu_{FE} C_{OX} (V_{GS} - V_{th})^2 \quad (2)$$

where I_{DS} is the drain-source current, W/L is the channel width-to-length ratio, C_{OX} is the gate dielectric capacitance per unit area, and V_{GS} and V_{th} represent the gate-source voltage and threshold voltage, respectively. To investigate the operational behavior of the Al-capped ZTO TFTs, TCAD simulation was conducted using ATHENA and ATLAS simulators (Silvaco Inc.). The simulation parameters, including bandgap and mobility of amorphous ZTO (a-ZTO) and crystalline ZTO (c-ZTO), are listed in Supplementary Table 2 of the Supporting Information. A positive bias stress test was carried out at $V_{GS} = 20$ V and $V_{DS} = 0.1$ V.

Data availability

The data that support the findings of this study are available from the corresponding author, Sung Kyu Park (https://skpark@cau.ac.kr), upon reasonable request.

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Author contributions

D.N., S.-P.J., and S.K.P. initiated the concept of the study. S.K.P. and Y.-H.K. planned the materials and experimental design. D.N. and S.-P.J. carried out the fabrication and evaluation of ZTO TFTs, both with and without an Al capping layer. D.H.K., D.K., and Y.K. contributed to the materials development and device analysis. D.N. performed the TCAD simulation. D.N., S.-P.J., and S.K.P. drafted the manuscript, incorporating feedback from all authors. S.K.P. and Y.-H.K. revised the manuscript. All authors participated in discussions of the findings and approved the completed work.

Competing interests

The authors declare no competing interests.

Additional information

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