

Study of Local Band Bending in n-Channel In₂O₃ Thin-Film Transistors Under Gate and Drain Voltage Stress Using *Operando* Hard X-ray Photoelectron Spectroscopy

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ABSTRACT: In this study, we analyze In₂O₃ thin-film transistors (In₂O₃-TFT) using synchrotron-based hard X-ray photoelectron spectroscopy (HAXPES) in *operando* conditions. A bottom-gate In₂O₃-TFT with a high-*k* Al₂O₃ gate dielectric, grown on thermally oxidized silicon (SiO₂/p⁺-Si), was examined while operating at varying V_{GS} and V_{DS} . The results reveal that the In 3d core level binding energy varies along the horizontal channel length, driven by the potential gradient induced by V_{DS} . Furthermore, the polarity of V_{GS} under positive V_{DS} significantly affects the channel structure, and V_{DS} influences the vertical electrostatic potential in the buried gate dielectric and gate electrode. Analysis suggests that the primary source of drain-source leakage current (I_{DS}) arises from high intrinsic electron and defect/trap levels associated with oxygen vacancies (V_O^{2+}). Additionally, investigation beneath the Au/Ti drain and source electrodes under different bias voltages reveals substantial In₂O₃ reduction, leading to significant indium metal (In⁰) formation. Schematic models of carrier transport mechanisms are proposed to explain these findings.

Keywords: Indium oxide, Aluminum oxide, photoelectron spectroscopy, band bending, chemical structures, surface and interfaces.

I. INTRODUCTION

Thin-film transistor (TFT) devices are recognized as crucial to the advancement of modern flexible electronics, enabling groundbreaking applications and transformative technologies. The choice of semiconductor material in TFT devices plays a pivotal role, as it not only dictates manufacturing parameters, such as substrate compatibility and fabrication processes, but also profoundly impacts the mechanical properties (e.g., flexibility) and critical electrical characteristics (e.g., carrier mobility, on/off current ratio, threshold voltage, and subthreshold swing).

Amorphous silicon (a-Si) has been widely investigated for flexible electronic applications; however, its application is limited by low carrier mobility (<1 cm²/Vs). To address this, Polycrystalline silicon (poly-Si) has emerged as a high-performance alternative, offering superior mobility (>80 cm²/Vs) and excellent stability. Nonetheless, its practical application is limited by the high processing temperatures required (>450 °C) and the costly crystallization methods. Furthermore, interest in Si-based TFTs for transparent and flexible electronics has waned due to their relatively narrow band gap. In contrast, amorphous oxide semiconductors (AOS) with post-transition-metal cations have gained significant attention as a promising alternative. AOS materials exhibit degenerate band conduction and high carrier mobility (>10 cm²/Vs), fundamentally differing from the covalent bonding of traditional semiconductors^{1,2}. Their exceptional performance arises from the large overlap of *ns* orbitals in the $(n-1)d^{10}ns^0$ electronic configuration ($n \geq 4$), enabling high conductivity, superior transparency in the visible spectrum, and excellent mechanical flexibility³. AOS-based TFTs demonstrate carrier mobilities 20 to 40 times higher than those

of a-Si TFTs and can be processed at significantly lower temperatures. This low-temperature processing capability makes AOS materials ideally suited for integration onto plastic substrates, enabling flexible electronic applications without compromising their outstanding properties⁴.

AOS Indium oxide (In₂O₃) has emerged as a highly promising material for flexible electronics due to its low growth temperature, wide band gap⁵⁻⁷, and high n-type carrier mobility (~40 cm²/Vs), derived from the single free-electron-like conduction band associated with the In 5s orbital. However, its electrical properties are strongly influenced by processing temperature, primarily due to the presence of uncontrolled oxygen vacancies (V_O^{2+})⁸, which induce unintentional n-type conductivity and disrupt the stoichiometric balance. At processing temperatures above 200 °C, In₂O₃ exhibits behavior consistent with a transparent conductive oxide, characterized by low electrical resistivity (~10⁻⁴ Ωcm) and high optical transparency (>85 %). This high conductivity is primarily attributed to elevated carrier concentrations resulting from enhanced crystallinity and molecular orbital ordering. Conversely, at deposition temperatures below 150 °C, In₂O₃ functions as a transparent semiconducting oxide, demonstrating a saturation mobility exceeding 15 cm²/Vs and a threshold voltage near 0 V, making it well-suited for low-temperature processing and flexible electronic applications.

Despite substantial progress in mitigating the detrimental effects of V_O^{2+} in In₂O₃, oxygen diffusion from In₂O₃ thin films at room temperature and atmospheric pressure remains a persistent challenge, primarily due to the relatively low bond dissociation energy (BDE) of the In-O bond (346 kJ/mol⁹). The presence of V_O^{2+} results in unintentional n-type conductivity and stoichiometric imbalance, as each doubly charged V_O^{2+} donates two free electrons. This leads to a transition

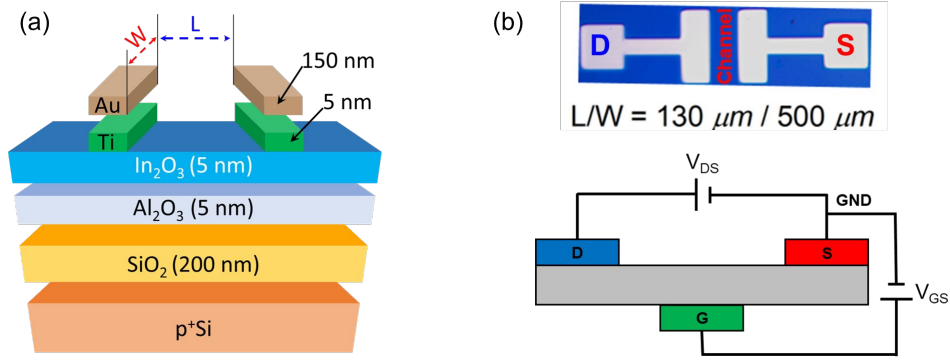


Figure 1: Schematic of the AOS-TFT device based on In_2O_3 : (a) featuring a bottom-gate and top-contact stacked layers. L and W represent the Length and Width of the channel, respectively; (b) electrical setup depicting a bias voltage applied at the gate (V_{GS}) to control the current from the source to the drain when applied (V_{DS}).

from semiconducting to metallic behavior with an increasing concentration of V_O^{2+} ^{10–12}. Consequently, as-deposited In_2O_3 often exhibits metallic-like conduction. Practically, the distribution of V_O^{2+} along the horizontal source-drain channel length direction and vertical semiconductor-gate oxide-gate semiconductor interface significantly influence the internal voltage drop, as well as the output and transfer characteristics of TFTs. For semiconductor use, precise control of carrier type and concentration across several orders of magnitude is essential for consistent device performance. Various strategies have been developed to address oxygen deficiencies, including annealing in oxygen-rich environments, exposure to N_2O plasma, or doping with metal cations that exhibit a high affinity for oxygen.

Given the critical influence of the channel's chemical structure on the stability and reliability of TFTs, achieving transformative advancements in In_2O_3 -TFTs necessitates the implementation of sophisticated and precise characterization techniques. Such advancements are essential to bridge the gap between fundamental research and industrial applications. To this end, the investigation of In_2O_3 -TFTs under actual operating conditions is paramount. These investigations require specialized experimental tools capable of providing detailed chemical and structural insights. Soft X-ray photoelectron spectroscopy (Soft-XPS) offers a non-destructive method for surface chemical analysis; however, its effectiveness is often limited to surface-level information, with insufficient access to bulk and buried interfaces. Conversely, hard X-ray photoelectron spectroscopy (HAXPES), leveraging high-energy synchrotron X-ray sources, enables non-destructive analysis of materials to depths of several tens of nanometers.

Here, we present an advanced *operando*-HAXPES analysis of the chemical dynamics in In_2O_3 -based TFT fabricated using the atomic layer deposition (ALD) technique. Unlike the channel lengths typically reported in the literature ranging from a few micrometers to submicrometer or nanometer scales, this investigation focuses on a channel length of several tens of micrometers. This choice was made to accommodate the micrometric size of the X-ray analysis beam, enabling precise investigation of the lateral channel structure, including the drain, center, and source regions, under applied bias volt-

age. The analysis demonstrates a clear correlation between the chemical structure and the current-voltage characteristics of the In_2O_3 -TFT, providing valuable insights into its operational behavior.

II. EXPERIMENTAL METHODS

A. TFT fabrication

This study investigates an In_2O_3 -TFT with a bottom-gate/top-contact configuration, as illustrated in Figure 1 (a). The device fabrication involved the deposition of a 5-nm-thick In_2O_3 layer atop a 5-nm-thick Al_2O_3 layer using ALD on heavily p-type doped silicon ($\text{p}^+\text{-Si}$) substrate serving as the gate electrode. The substrate included a 200-nm-thick thermally grown SiO_2 dielectric layer. Before deposition, the $\text{SiO}_2/\text{p}^+\text{-Si}$ substrate underwent a rigorous cleaning process, including ultrasonic cleaning in ethanol and acetone, ultraviolet/ozone treatment for 20 minutes, and a final rinse in a 4.8 % HF solution to ensure optimal surface preparation¹³.

The Al_2O_3 layer was deposited via ALD using trimethylaluminum ($(\text{CH}_3)_3\text{Al}$) as the precursor and O_3 as the oxidant gas, at a substrate temperature of 200 °C. Subsequently, the In_2O_3 layer was grown atop the Al_2O_3 layer at 150 °C using an ALD process that employed the ethylcyclopentadienyl indium (InEtCp) precursor and a combination of H_2O and O_3 as oxidant gases. The InEtCp precursor temperature was maintained at 80 °C, with N_2 utilized as both the carrier and purge gas to minimize unwanted gas-phase reactions.

To enhance the properties of the indium oxide, post-thermal annealing (PTA) process was performed under an O_2 atmosphere containing a maximum of 5% O_3 at 150 °C for 90 minutes. To complete the TFT structure, 5/150 nm thick Ti/Au ohmic contact layers were patterned onto the In_2O_3 layer using electron beam evaporation through a shadow mask, forming the source and drain electrodes. Finally, Figure 1 (b) shows the associated electrical setup as well as the actual channel dimensions determined through optical measurements (width (W) of 500 μm and a length (L) of 130 μm).

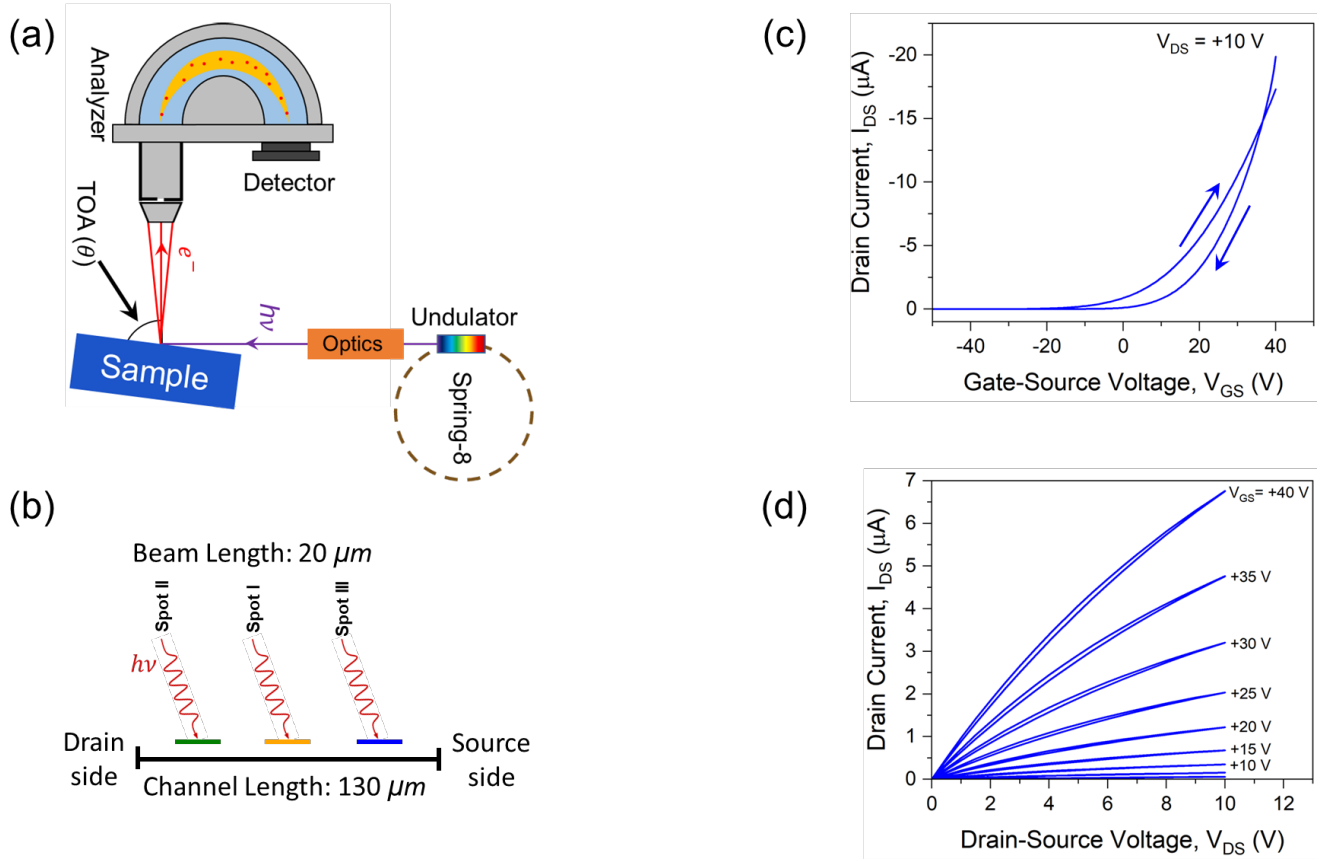


Figure 2: Schematic illustration of the probed In₂O₃-TFT device: (a) measurement setup of HAXPES-based on Synchrotron X-ray source, (b) Three distinct spots (I, II, III) on the In₂O₃ channel surface probed by the X-ray beam, (c) Transfer characteristics for a drain voltage (V_{DS}) of +10 V, (d) Output characteristics for various gate voltages (V_{GS} ranging from 0 V to +40 V).

B. TFT characterization

The HAXPES analysis of the In₂O₃-TFT sample was performed using synchrotron radiation at beamline BL09XU, SPring-8 (Japan)^{14–16}. To accurately replicate the operating conditions of TFT, which involve substantial current flow through the channel in response to applied gate and drain voltages, an electric manipulator capable of connecting at least three terminals is required. The HAXPES tool from EH2 at BL09XU was selected for this study due to its four-terminal connectivity. The experimental setup for *operando*-HAXPES is shown in Figure 2 (a). All measurements were conducted at room temperature under a stable base pressure of approximately 10^{-5} Pa. Photon excitation was set to an energy of 5.95 keV, and the emitted photoelectrons were analyzed using a hemispherical electron analyzer (VG-Scienta R4000) in angle-integrated transmission mode, with a $\pm 32^\circ$ acceptance angle. The incident angle of the X-ray was set at 35° to the sample surface. At that time, the take-off angle (TOA) of the photoelectrons was 23° to 87° . The total energy resolution, accounting for both photon and spectrometer bandwidths, was determined to be 0.24 eV by fitting the Au Fermi edge. With a TOA of 55° , the X-ray beam projected a footprint of ap-

proximately 20 μm along the channel length, allowing distinct probing of three regions (drain, middle, and source) within the In₂O₃ channel, as illustrated in Figure 2 (b).

The binding energy (BE) calibration was performed using the Fermi edge of an evaporated Au film. The small beam size at BL09XU enabled precise lateral mapping of the micro-metric channel length of the TFT device. Theoretical inelastic mean free paths (IMFPs), calculated using the QUASES-IMFP-TPP2M software¹⁷, were employed to estimate the sampling depth (d), or escape depth of photoelectrons, as $d \sim 3 \times \text{IMFP} \times \sin(\text{TOA})$, yielding a value of approximately 30 nm for the 5.95 keV photon energy used. HAXPES data analysis was primarily conducted using CasaXPS software (CASA Software Ltd.¹⁸). Core level spectra were fitted using a pseudo-Voigt peak shape (a combination of Gaussian and Lorentzian functions), with background subtraction performed using the Shirley algorithm. The Lorentzian contributions to the investigated core levels were found to be minimal and were maintained at a fixed width during the fitting of each core level.

During the *operando*-HAXPES analysis, two ADCMT 6241A DC voltage-current source/monitor units were employed to apply a continuous bias voltage between the Gate-

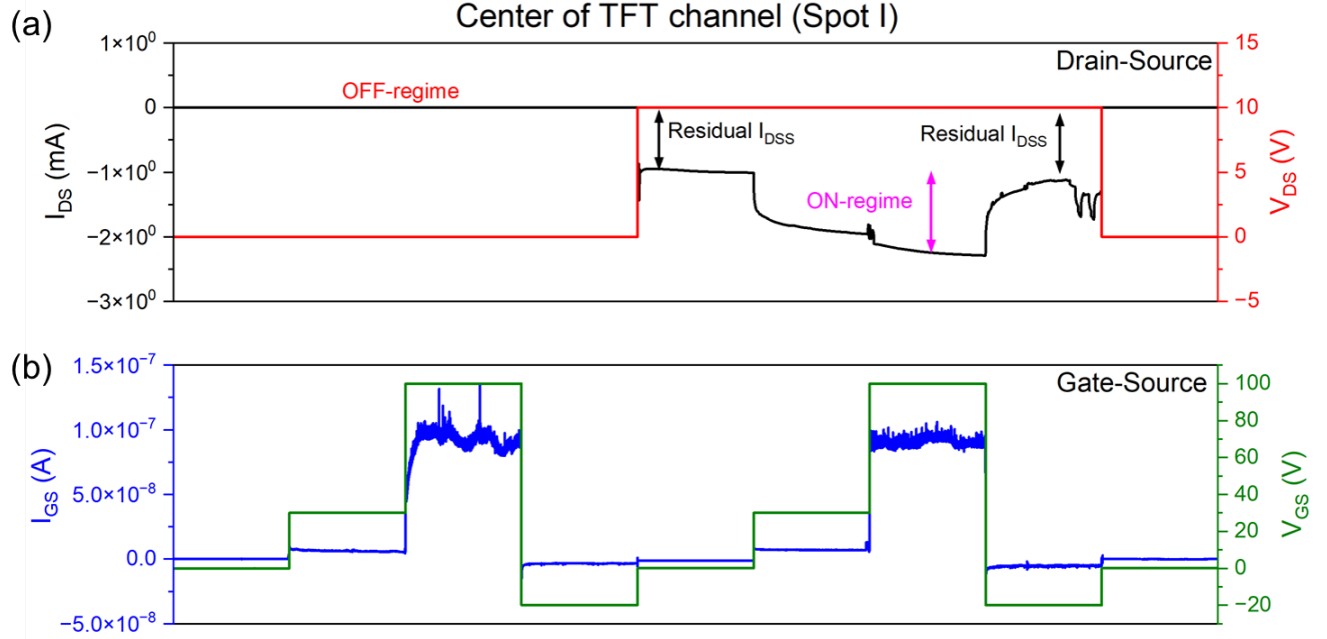


Figure 3: Current-voltage measurements during the *operando*-HAXPES analysis at spot I (center of the TFT channel): (a) I_{DS} -time curve for varying drain-source ($V_{DS} = 0$ V and 10 V) voltages, (b) I_{GS} -time curve for different gate-source ($V_{GS} = 0$ V, 30 V, 100 V, and -20 V) voltages.

Source (V_{GS}) and Drain-Source (V_{DS}) terminals. In addition, the current-voltage (transfer/output) characteristics of the In_2O_3 -TFT were measured in atmospheric environment using a B1500A semiconductor device analyzer (Agilent Technologies) before the HAXPES measurements to verify the device's operational status, as shown in Figure 2 (c-d).

III. RESULTS AND DISCUSSION

Fundamentally, the application of a positive V_{GS} bias to the TFT (before the *operando*-HAXPES analysis) leads to the accumulation and confinement of electrons at the interface between the In_2O_3 semiconductor layer and the Al_2O_3 gate dielectric, resulting in the formation of a conductive channel. When a bias is applied between the source and drain ($V_{DS} > 0$ V), the accumulated carriers drift from the source to the drain terminal, generating a drain-source current. This current increases with higher V_{DS} and V_{GS} values, eventually reaching saturation at sufficiently large V_{GS} . The variation of the current with changing V_{DS} and V_{GS} in a conventional TFT is illustrated in Figure 2 (c) and Figure 2 (d), which present the representative transfer and output characteristics measured for the In_2O_3 -TFT, respectively. The output characteristic is defined by the drain-source current (I_{DS}) as a function of V_{DS} at fixed V_{GS} values. Conversely, the transfer characteristic is determined by measuring I_{DS} as a function of V_{GS} at fixed V_{DS} . In such devices, the channel conductance (between the drain and source electrodes) is governed by the density of free electrons (for n-channel devices) within the transistor channel. This charge density can be controlled by applying a V_{GS}

to the gate electrode.

In Figure 2 (c), V_{GS} was swept from -45 V to 40 V and then back to -45 V in 0.5 V increments, with a fixed V_{DS} of 10 V. Despite a macroscopic channel size, the In_2O_3 -TFT exhibits typical n-channel transistor characteristics with reasonable ON/OFF behavior. The ON-current (I_{DS}) increases with rising V_{GS} , demonstrating gate voltage modulation, although current saturation was not observed, as reported in previous studies^{19,20}. Additionally, the device exhibits large negative values for both the turn-on voltage (V_{ON}) and threshold voltage (V_{th}), with the drain current reaching approximately -20 μA at $V_{GS} = 40$ V. The transfer curves reveal significant hysteresis, with the backward sweep currents being lower than the forward sweep currents. This behavior is likely caused by charge carrier trapping within the In_2O_3 semiconductor film, leading to variations in the threshold voltage. The output characteristics, presented in Figure 2 (d), similarly display a lack of I_{DS} saturation at $V_{DS} = 10$ V, with the device remaining in the ohmic region for all applied V_{GS} values (0 V to +40 V). This absence of current saturation may be attributed to the pronounced n-type characteristics of In_2O_3 , including a high density of trap states or defects at the channel/dielectric interface or within the dielectric layer itself^{21,22}.

The transfer and output characteristics confirm the switching operation of the In_2O_3 -TFT, demonstrating its ability to transition between the OFF (cut-off region) and ON states. Furthermore, the device exhibits effective voltage control functionality, wherein the gate-to-source voltage (V_{GS}) modulates the current flow from the drain to the source. By applying and increasing V_{GS} , additional charge carriers are introduced into the channel, resulting in enhanced conductivity of

the In_2O_3 channel.

Figure 3 depicts the current-voltage (I-V) characteristics of the In_2O_3 -TFT recorded during chemical analysis at the central region (spot I) along the channel length. Figure 3(a) presents the drain-source current (I_{DS}) as a function of V_{DS} while sweeping V_{GS} at 0 V, 30 V, 100 V, and -20 V. Similarly, Figure 3(b) shows the gate-source current (I_{GS}) as a function of V_{GS} , with V_{DS} fixed at 0 V and +10 V. Unlike the dynamic-mode electrical characteristics shown in Figure 2(c-d), where applied voltages were continuously varied to demonstrate the ON/OFF switching behavior of the device, the I-V data in Figure 3 were obtained under static conditions, with fixed voltages applied over time for each measurement step. This static characterization approach facilitates the chemical investigation of the In_2O_3 film, as well as the buried layers and interfaces, under varying bias voltage conditions.

I_{GS} current exhibits a proportional increase or decrease with corresponding increases or decreases in V_{GS} , respectively while remaining independent of the V_{DS} value. I_{GS} current through the Al_2O_3 dielectric gate layer is relatively low, ranging from approximately -5×10^{-9} A at $V_{GS} = -20$ V to approximately 9×10^{-8} A at $V_{GS} = 100$ V. For the I_{DS} current, it is confirmed that no significant current flows from the drain to the source at $V_{DS} = 0$ V, indicating that the transistor is in the OFF state as expected. However, at $V_{DS} = 10$ V, a notable residual I_{DS} current is observed when V_{GS} is set to 0 V or -20 V.

This residual current, referred to as the drain-source leakage current (I_{DSS}), represents the current flow from the drain to the source terminal when the transistor is in the OFF state ($V_{GS} = 0$ V). Such leakage current contributes to increased power loss and reduced device efficiency. The magnitude of I_{DSS} is typically measured under specific conditions, defined by particular values of V_{GS} and V_{DS} . Under ultra-high vacuum (UHV) conditions, the large observed I_{DSS} is likely induced by the lack of oxygen from In_2O_3 , especially when using a macroscopic channel size. However, unlike the high I_{DSS} values ($\sim$ mA) recorded in Figure 3(a), the I_{DSS} observed in Figure 2(c) and Figure 2(d) under atmospheric pressure and OFF-state conditions are nearly 0 μA for the same device.

The origin of this significant I_{DSS} value can be attributed to oxygen reduction as well as intrinsic defects and traps within the In_2O_3 semiconducting layer. In the absence of a deposited protective capping layer on the channel surface, UHV conditions (during the *operando*-HAXPES analysis) can exacerbate this effect by removing released oxygen and preventing replacement from external sources, resulting in a generation of more defects. Based on the relatively stable I_{DSS} at $V_{DS} = 10$ V and $V_{GS} = 0$ V / -20 V, we assume that the formation of oxygen vacancies in the In_2O_3 channel reaches a stable or saturation point where further oxygen vacancy defects creation is unfavorable.

This process leads to the reduction of oxide films and subsequent changes in their chemical and electronic structures, such as the emergence of metallic states. These changes may directly impact the electrical behavior, including the I_{DSS} . Furthermore, the actual I_{DS} indicative of the ON state is observed when the transistor operates at V_{GS} values of 30 V and 100 V.

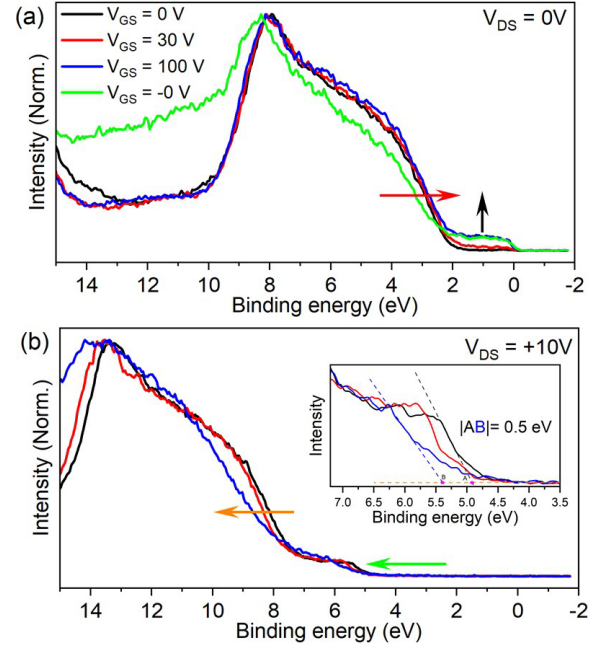


Figure 4: Valence-band spectra from In_2O_3 -TFT under different V_{GS} : (a) $V_{DS} = 0\text{V}$ with the black arrow showing the increase of the reduced indium and the red arrow showing the shift at lower banding energy of the edge of the valence band and (b) $V_{DS} = 10\text{V}$ with both the green and orange arrows showing a shift at lower banding energy of reduced indium and the edge of the valence. The inset in (b) also report the energy shift of the reduced indium under different V_{GS} at $V_{DS} = 10\text{V}$.

Similar I-V characteristics obtained during the analysis of the drain and source regions are presented in ESI, p. S-3, Figures S2 and S3, respectively. To gain deeper insights into these underlying mechanisms, the chemical changes occurring under applied bias voltage will be thoroughly analyzed and discussed.

Figure 4 presents valence band spectra obtained as a function of V_{DS} and V_{GS} at spot I (Figure 2(b)), corresponding to the center of the In_2O_3 channel. We note that the valence band top on the low energy side is solely derived from In_2O_3 (not underneath Al_2O_3 or SiO_2 contributions). The data shown in Figure 4(a) seems indicated that there is a progressive reduction of the In_2O_3 valence band edge with increasing V_{GS} , as highlight with the black arrow spanning the energy range from 0 eV (Fermi level) to 2 eV (valence band edge). Indeed, with an increase in the applied voltage, electron density is generated in the in-gap region, filling the Fermi level with electrons and exhibiting metallic behavior. This may be due to electron accumulation and oxygen defect formation leading to the occupation of the lowest empty conduction-band states. We also observe that the electron density still remains (green curve) even when the applied voltage is removed ($V_{GS} = -0$ V). At $V_{DS} = 0$ V, the intensity of the valence band edge reaches its maximum at $V_{GS} = 100$ V (blue curve) and does not revert to its initial state upon returning to $V_{GS} = -0$ V (green

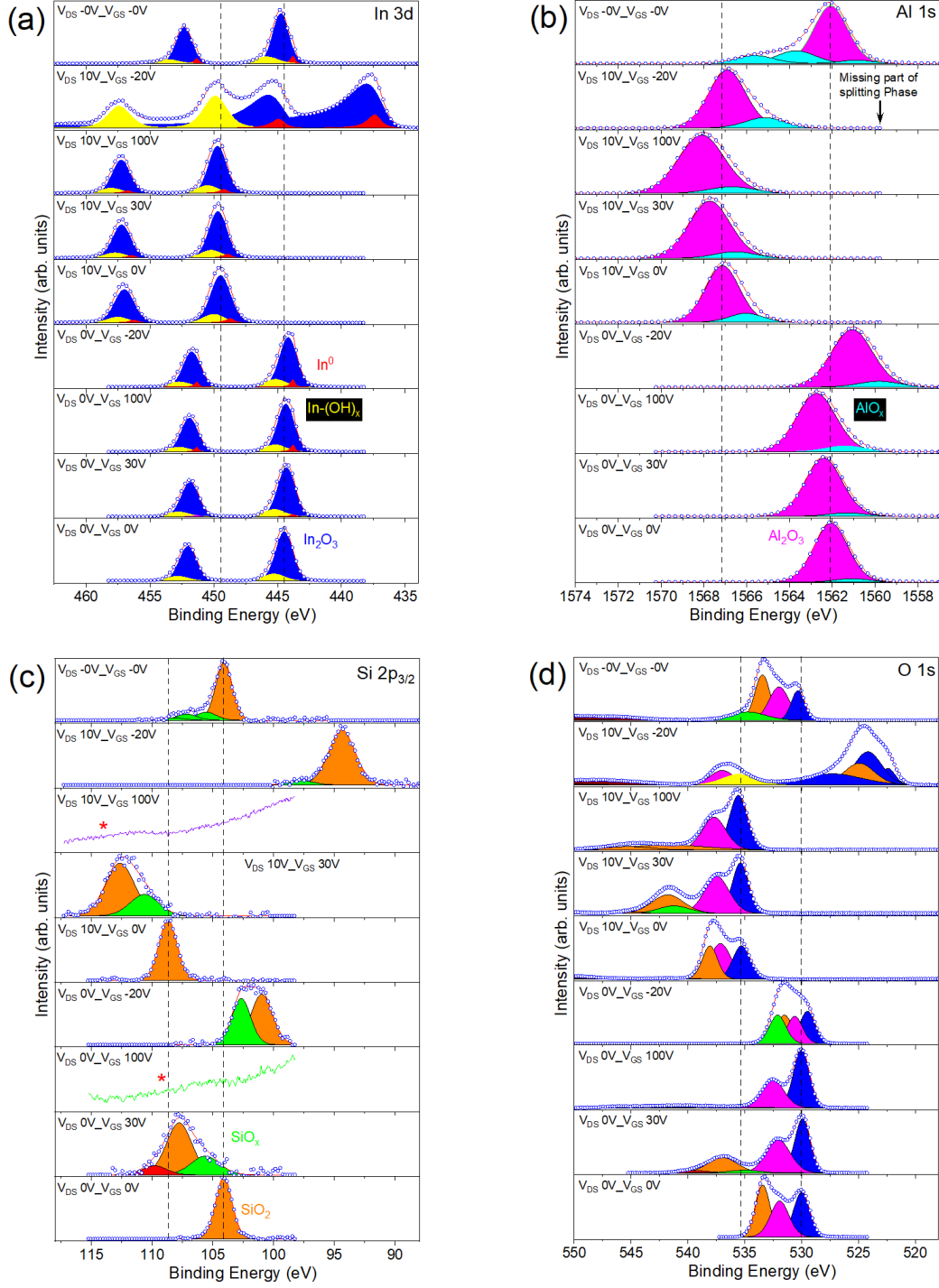


Figure 5: Core level HAXPES spectra from In_2O_3 -TFT under constantly applying V_{DS} and V_{GS} : (a) In 3d, (b) Al 1s, (c) Si 2p_{3/2} and (d) O 1s

curve). It should be noted that, although a minor contribution from beam-induced effects during prolonged HAXPES exposure cannot be entirely excluded, several experimental observations, namely the absence of burn marks at the beam spot, the stability of the UHV pressure level

over time, and the lack of spectral drift or broadening during measurement indicate that sample degradation due to X-ray radiation is unlikely. In addition, it is worth noting that the area (spanning from 0 eV to 2 eV) under the valence band edge of blue and green curves, relative to the total area

TABLE I: Summary of the peak fitting parameters from In 3d_{5/2}, Al 1s, Si 2p_{3/2} and O 1s elements: colors, components and binding energy (BE) in eV as the function of the applied V_{DS} and V_{GS} bias voltage are respectively listed. # corresponds to yellow peak.

Elements	Colors	Components	Binding energy (eV) as function of the V_{DS} / V_{GS} (Volt / Volt)								
			0 / 0	0 / 30	0 / 100	0 / -20	10 / 0	10 / 30	10 / 100	10 / -20	-0 / -0
In 3d _{5/2}	Red	In ⁰	-	443.8	443.8	443.8	448.7	448.9	449.2	437.4	443.8
	Blue	In ₂ O ₃	444.5	444.3	444.3	444.1	449.5	449.7	449.7	437.4	444.8
	Yellow	In-(OH) _x	445.3	445.3	445.2	445.2	450.0	450.2	450.5	449.9	445.9
Al 1s	Magenta	Al ₂ O ₃	1562.1	1562.4	1562.7	1561.1	1567.1	1567.7	1568.1	1566.9(*)	1562.1
	Cyan	AlO _x	1561.1	1561.3	1561.3	1559.8	1566.0	1566.5	1566.7	1565.2	1565.6/1563.7/1560.9
Si 2p _{3/2}	Orange	SiO ₂	104.1	107.8	~108.0	101.0	108.7	112.6	~116.0	94.4	104.1
	Green	SiO _x	-	105.6	~106.0	102.7	-	110.6	~111.0	97.8	108.2/107.2/105.6
O 1s	Blue	In ₂ O ₃	530.0	529.9	530.0	529.5	535.3	535.4	535.6	535.5(#)/527.3/524.2/522.4	
	Magenta	Al ₂ O ₃	531.9	532.0	532.5	530.6	537.1	537.5	537.7	536.9	532.0
	Orange	SiO ₂	533.4	537.0	545.5/541.0	532.1	538.1	541.8	541.8	524.9	533.4
	Green	SiO _x	-	535.0	536.9	531.5	-	541.2	545.3	~528.3	534.7

(from 0 eV to 12 eV) of the valence band structure, is approximately 4 %. Furthermore, as V_{GS} increases from 0 V to 100 V, the position of the valence band maximum (VBM) shifts towards lower binding energies, as denoted by the red arrow. For $V_{GS} = -0$ V, the position of the maximum intensity of the green curve (located around 8 eV) slightly shifts around 0.4 eV towards higher binding energy compare to the maximum intensity of black curve ($V_{GS} = 0$ V).

In Figure 4(b), under $V_{DS} = 10$ V, a global shift of the valence band structure to higher binding energies (approximately 5 eV) relative to the zero-binding-energy position is noted, as marked by the green arrow. We also observe a high-binding-energy shift of the VBM is observed as V_{GS} increases from 0 V to 100 V, as indicated by the orange arrow. Then, it is noteworthy that at $V_{DS} = 0$ V, the intensity of the valence band edge increases at a fixed energy position, whereas under $V_{DS} = 10$ V, the intensity of the valence band edge remains constant, and its energy position shifts depending on the value and polarity of V_{GS} as seen in the inset of Figure 4(b). This stable intensity at around 4 % is strongly consistent with the stable I_{DSS} value at $V_{DS} = 10$ V as already underlined in Figure 3(a).

Figure 5 presents the peak fittings obtained under various V_{DS} and V_{GS} conditions for (a) In 3d from In₂O₃, (b) Al 1s from Al₂O₃, (c) Si 2p_{3/2} from SiO₂, and (d) O 1s from the three aforementioned layers. Similar to the data shown in Figure 4, these measurements were performed at the spot I (Figure 2(b)), corresponding to the center of the channel. The binding energy positions and their respective components as a function of V_{DS} and V_{GS} are summarized in Table I. It is important to note that, with a photon energy of 5.95 keV, it is impossible to chemically resolve the chemical structure of the buried p⁺-Si gate electrode beneath the 200 nm SiO₂ layer. Therefore, we attribute all observed Si 1s signals exclusively to the SiO₂ dielectric layer.

For clarity, the intensity in Figure 5 is presented in arbitrary units to better highlight the different spectral components. In the case of the In3d core level, the actual integrated areas of

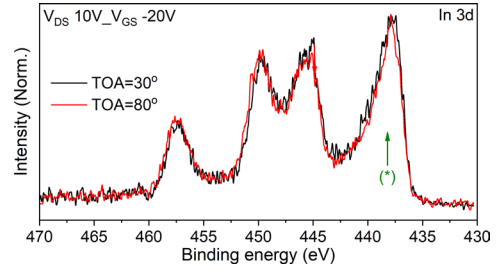


Figure 6: Angle dependent analysis of In 3d at $V_{DS} = 10$ V and $V_{GS} = -20$ V.

all spectra as a function of V_{GS} and V_{DS} are shown in ESI, p. S-4, Figures S5. The quantitative fitting of the In 3d core-level spectra, enabling the distinction between In³⁺ and In⁰ components and the monitoring of their evolution under varying bias conditions, is presented in ESI, p. S-4, Table S1. At $V_{DS} = 0$ V, three primary components of the In 3d core-level spectrum can be identified, as presented in Figure 5(a). The intensity of the In⁰ component (red peak) increases noticeably when a gate voltage ($V_{GS} \neq 0$ V) is applied to the sample, irrespective of the electric field direction in the adjacent Al₂O₃ dielectric layer. This observation aligns with the previously reported enhancement of the valence band edge, as illustrated in Figure 4(a). Then, an energy shift of the In₂O₃ component (blue peak) toward lower binding energies is observed, correlating with the increase in In⁰ intensity. Notably, the direction of the In₂O₃ energy shift remains independent of V_{GS} . Lastly, the In-(OH)_x component (yellow peak) also exhibits a shift to lower binding energies under V_{GS} . The presence of hydroxyl (-OH) groups in the In₂O₃ thin film is linked to its fabrication at a relatively low ALD deposition temperature of 200°C. Indeed, the use of H₂O as oxidant likely leads to residual -OH groups due to a self-limiting growth which likely lead to residual -OH groups if not fully reacted in the

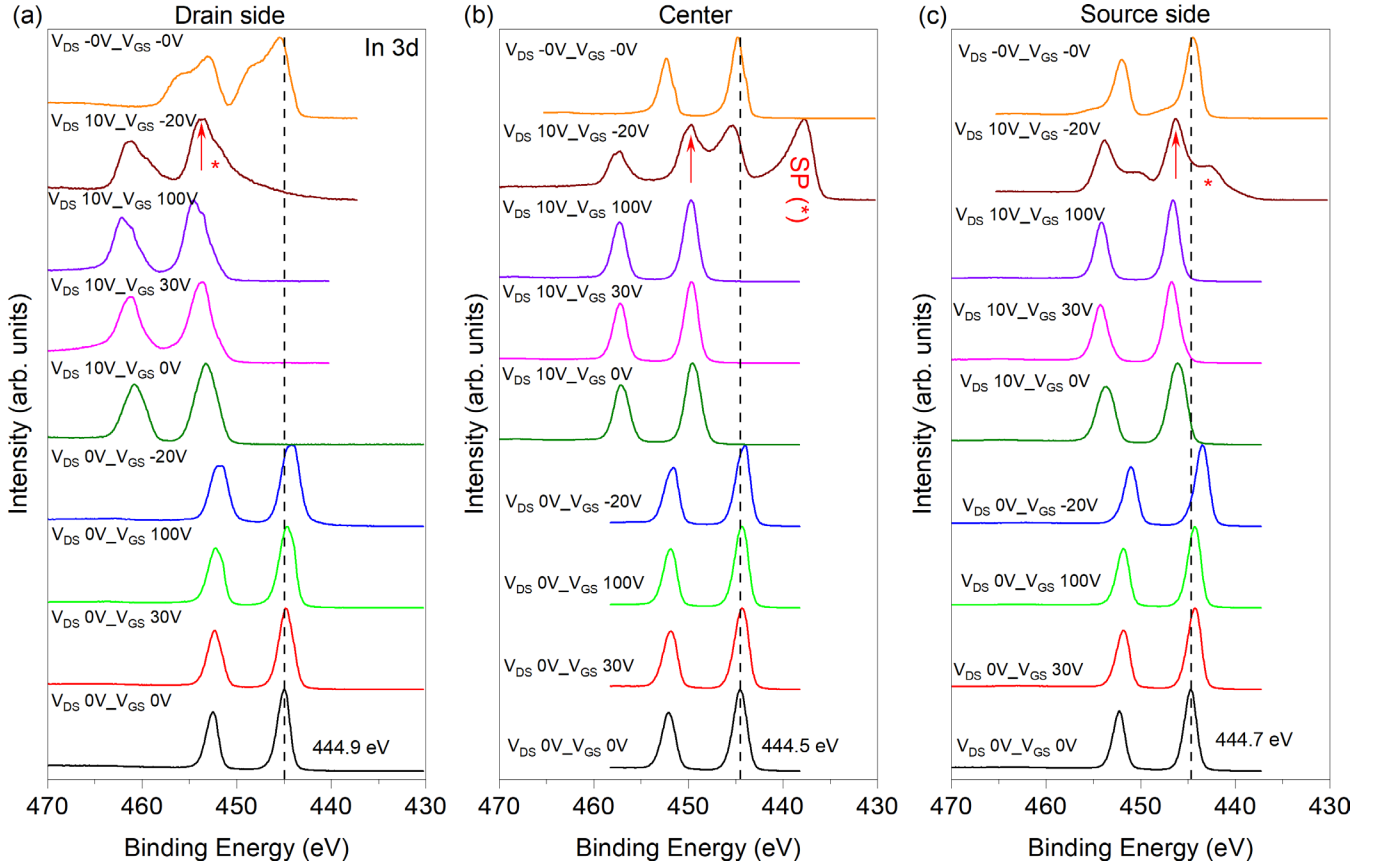


Figure 7: In 3d core level shifts under varying V_{GS} and V_{DS} : (a) drain side from spot II, (b) center area from spot I, and (c) source side from spot III.

next cycle. Thus, the low temperature will prevent complete desorption of byproducts from the incomplete reaction. Furthermore, based on the peak fitting at $V_{DS} = 0$ V and $V_{GS} = 0$ V, the intensity of the yellow component corresponding to $\text{In}(\text{OH})_x$ was fixed at 15 % of the total In 3d peak intensity, with the combined intensity of the In_2O_3 and In^0 components comprising the remaining 85 %.

Under similar experimental conditions (at $V_{DS} = 0$ V with varying V_{GS}), the Al 1s and Si $2p_{3/2}$ core levels, as presented in Figure 5 (b-c), exhibit systematic binding energy shifts that depend on both the polarity and magnitude of V_{GS} . Notably, the oxide films (Al_2O_3 and SiO_2) function as insulating barriers, impeding direct electron transport between the source and gate while permitting the applied electric field (V_{GS}) to modulate charge carrier flow within the channel. As illustrated in Figure 5 (b), the binding energies of Al_2O_3 (magenta) and AlO_x (cyan) shift towards higher values at $V_{GS} = 30$ V and 100 V relative to $V_{GS} = 0$ V, whereas a shift towards lower binding energy is observed at $V_{GS} = -20$ V. Furthermore, analysis of the Si $2p_{3/2}$ spectra reveals that the application of bias voltage to the bottom gate facilitates the dissociation of silicon oxide into distinct SiO_2 (orange) and SiO_x (green) phases, as evidenced by the emergence of phase splitting at $V_{GS} = 30$ V and -20 V. At $V_{GS} = 100$ V, the Si $2p_{3/2}$ core level exhibits broadening towards higher binding energies, likely attributable to

the increased applied voltage. This splitting and broadening of the Si signal is further corroborated by the Si 2s spectra (refer to ESI, p. S-4, Figure S4).

To further examine the influence of V_{DS} on the chemical structure of the TFT, a horizontal bias of 10 V was applied across the channel, between the source and drain electrodes. Compared to the spectra recorded under unbiased conditions ($V_{DS} = 0$ V and $V_{GS} = 0$ V), the application of $V_{DS} = 10$ V at $V_{GS} = 0$ V induces an approximately 5 eV shift of the In_2O_3 (blue peak) towards higher binding energies. Moreover, this blue component undergoes a further, albeit slight, shift to higher binding energies when V_{GS} is increased to 30 V and 100 V under $V_{DS} = 10$ V, in contrast to the binding energy shifts observed at $V_{DS} = 0$ V. In the case of the In^0 (red peak) and $\text{In}(\text{OH})_x$ (yellow peak) components, no significant changes in their relative intensities are detected as V_{GS} is varied from 0 V to 30 V and 100 V. However, their binding energy positions exhibit a dependence on V_{GS} similar to that of the primary blue component. Notably, the intensity of the In^0 signal remains unchanged upon the application of $V_{DS} = 10$ V, maintaining the same intensity as that observed at $V_{DS} = 0$ V and $V_{GS} = -20$ V. Furthermore, at $V_{GS} = -20$ V, the In_2O_3 signal undergoes a pronounced shift towards lower binding energies, accompanied by the emergence of an extended low-binding-energy tail an effect absent under $V_{DS} =$

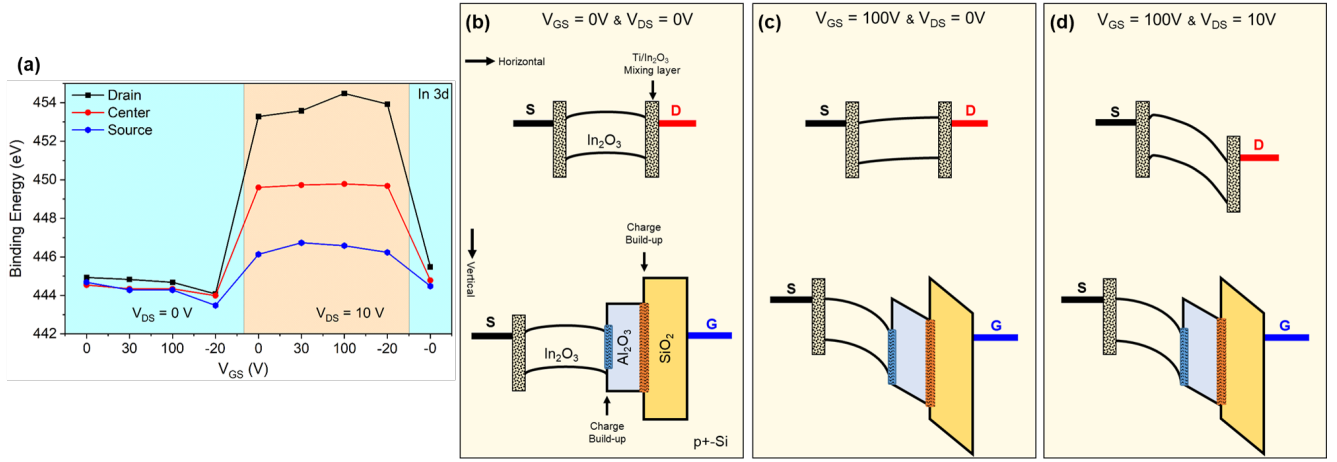


Figure 8: (a) Binding energy shift of In 3d_{5/2} core levels upon V_{DS} and V_{GS} bias voltages; (b-d) Conceptual band diagrams for the horizontal source-drain (Au/Ti/In₂O₃/Ti/Au) current path at the top and the vertical current control structure between the source-gate (Au/Ti/In₂O₃/Al₂O₃/SiO₂/p⁺-Si) at the bottom under different combination of V_{DS} and V_{GS} bias voltages.

0 V. Since the relative intensities of the red, blue, and yellow peaks were constrained during peak fitting, it is evident that In-(OH)_x does not undergo a significant shift towards lower binding energies, unlike In⁰ and In₂O₃.

Consistent with the behavior observed for the In 3d core level, both Al 1s and Si 2p_{3/2} exhibit a global binding energy shift of approximately 5 eV when transitioning from V_{DS} = 0 V and V_{GS} = 0 V to V_{DS} = 10 V and V_{GS} = 0 V. This observation is particularly intriguing, given that the applied V_{DS} = 10 V is directed laterally along the channel layer rather than vertically across the channel and the p⁺-Si substrate. The underlying mechanism responsible for this unexpected result will be examined in detail in a subsequent section of this manuscript. In addition to this global shift, a similar trend is observed for Al 1s and Si 2p_{3/2} at V_{DS} = 10 V with V_{GS} = 30 V and 100 V, compared to V_{DS} = 0 V under the same gate bias conditions. Notably, the Si 2p_{3/2} core level exhibits a pronounced shift towards lower binding energies, analogous to the shift observed in In 3d. Furthermore, due to the exploratory nature of this study and the unprecedented magnitude of energy splitting induced by biasing, a complete acquisition of the Al 1s core level was not carried out at V_{DS} = 10 V and V_{GS} = -20 V, as shown in Figure 5 (b). At V_{DS} = -0 V and V_{GS} = -0 V, the spectral profile of the In₂O₃ (blue component) is restored with a slight binding energy shift relative to the unbiased state, whereas no residual binding energy shift is observed for the Al 1s and Si 2p_{3/2} core levels. Additionally, the analysis of In 3d, Al 1s, and Si 2p_{3/2} is in agreement with the O 1s spectral data, which reveal a progressive shift of the O 1s blue peak associated with In₂O₃ towards lower binding energies, consistent with the trend observed for In 3d. The O 1s data further confirm the influence of both the polarity and magnitude of the applied bias voltage on Al₂O₃ (magenta) and SiO₂ (orange), as well as the pronounced energy splitting behavior of the SiO₂ layer. Finally, additional spectra of In 3d, Al 1s, Si 2p_{3/2}, and O 1s core levels were acquired from a similar In₂O₃-TFT featuring a 70 μm channel length, with

measurements taken near the center of the channel (ESI, p. S-3, Figure S1). As demonstrated in ESI, p. S-5, Figure S6, the recorded core level data are consistent with the results presented in Figure 5, thereby confirming the reproducibility of these measurements.

To elucidate the distribution of the In 3d splitting phase shifted at lower binding energy, the angle-resolved (angle-integrated transmission mode) data at V_{DS} = 10 V and V_{GS} = -20 V were compared as shown in Figure 6. These measurements were carried out with a fixed take-off angle of 50 °C, meaning that photoemission angles ranging from 30 °C (i.e. minimum sampling depth) to 80 °C (i.e. maximum sampling depth) were captured in a single shot. Comparing the maximum and minimum depth-resolved information in Figure 6, no significant differences related to the splitting phase (marked with an asterisk and green arrow) were observed. This finding indicates that the splitting phase is not localized to the In₂O₃ surface or near the In₂O₃/Al₂O₃ interface but is homogeneously distributed within the In₂O₃ layer.

In Figure 7, we present an extended In 3d core level analysis at various X-ray beam positions and under different V_{DS} and V_{GS} configurations. This position-dependent investigation (drain side, center, and source side) enables us to evaluate binding energy changes along the horizontal direction of channel, particularly under V_{DS} bias. Firstly, at V_{DS} = 0 V and V_{GS} = 0 V (black curves), we observe that the In 3d_{5/2} binding energy is lower around the center (444.5 eV) compared to the electrode side edges (source/drain: 444.7 eV/444.9 eV). These differences can be attributed to the Fermi level alignment process between the metal electrode (source/drain) and the semiconductor, which can cause band bending of the In₂O₃ in equilibrium in the case of imperfect contact on an actual device. Additionally, the position-dependent analysis confirms the progressive shift of In 3d towards lower binding energy at V_{DS} = 0 V and V_{GS} = 30 V, 100 V, and -20 V, with a similar displacement observed at all three positions. Furthermore, under V_{DS} = 10 V, we observe different residual

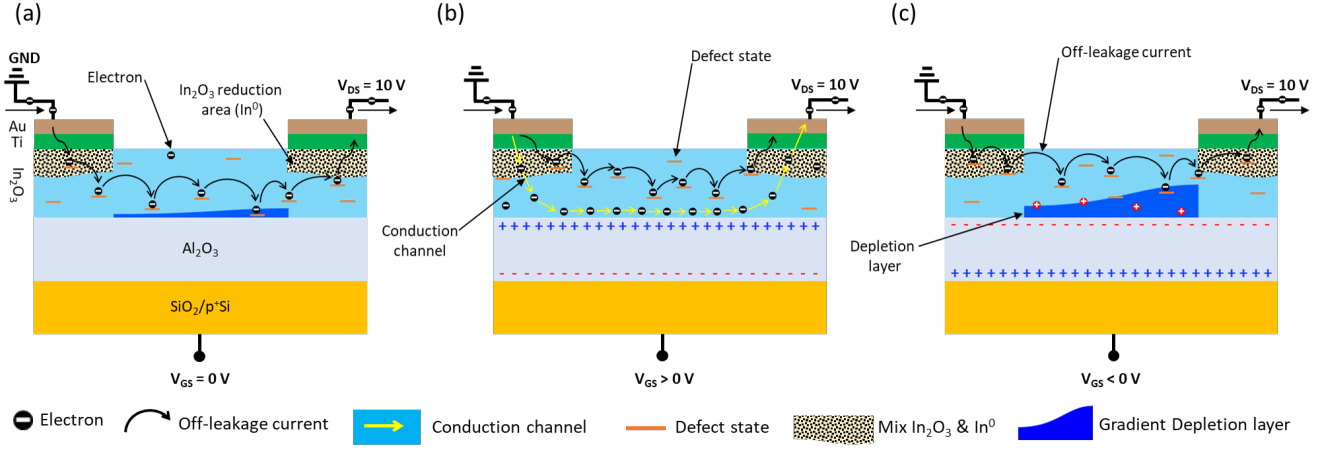


Figure 9: Schematic images for the transport of electron under different V_{GS} and V_{DS} biases

energy shifts from the source side (lower residual shift) to the drain side (higher residual shift) as seen at $V_{GS} = 0$ V (olive curves). This residual shift, relative to the binding energy of In 3d at $V_{DS} = 0$ V and $V_{GS} = 0$ V, is associated with the transverse voltage between the source and the drain. The source is grounded, so the quasi-Fermi level remains unchanged from the equilibrium, unlike the drain. The electrical potential difference across the channel induces a potential gradient that varies smoothly across the channel, leading to the observed binding energy shift. In addition to the residual shift at the drain, center, and source positions, we also validate the slight high binding energy shift observed earlier at $V_{DS} = 10$ V and $V_{GS} = 30$ V and 100 V in Figure 5.

When the gate is made negative ($V_{DS} = 10$ V and $V_{GS} = -20$ V) relative to the source, the In_2O_3 separation phase (SP) in wine curves is observed at all three positions (drain side, center, and source side) with varying intensities. This indicates a position-dependent driving force facilitating the separation. Notably, the intensity of the SP is higher at the center of the channel compared to the edges (drain/source terminals). This SP phase can be linked to the presence of a carrier depletion. The higher intensity of the SP at the center compared to the edges is likely associated to the shape of the depletion region at the center of In_2O_3 , which diminishes near the drain/source. This progressive reduction and eventual disappearance of the depletion zone near the channel terminals may correlate with the reduction process of iodine beneath the Ti interlayer as observed in ESI, p. S-7, Figure S10. At reverse $V_{DS} = -0$ V and reverse $V_{GS} = -0$ V, the In 3d peak positions only show a slight shift compare to the initial $V_{DS} = 0$ V and reverse $V_{GS} = 0$ V. It is noteworthy that, unlike the In 3d spectra from the center and source sides, the In 3d spectra from the drain side are broader and exhibit a shoulder peak, which may be a result of the previous biasing at $V_{DS} = 10$ V.

Figure 8(a) illustrates a clear dependence of the In $3d_{5/2}$ on both V_{DS} and V_{GS} applied bias values ($V_{DS} = 0$ V and $V_{GS} = 0$ V from Figure 7 being the binding energy reference). From Figure 8(a) we basically observe a significant energy shift when V_{DS} is applied. At $V_{DS} = 0$ V, a slight and pro-

gressive energy shift towards lower binding energy is noted, irrespective of the V_{GS} values. Under $V_{DS} = 10$ V, the In $3d_{5/2}$ shifts from the source, center, and drain probing positions follow a similar trend, although the extent of these shifts varies. This suggests that the energy shift is influenced by the potential gradient along the drain-source direction. Finally, we note that the position of the In $3d_{5/2}$ peak changes slightly with variations in gate voltage relative to $V_{DS} = 10$ V and $V_{GS} = 0$ V.

The bias-dependent binding energy evolution can be rationalized using schematic energy band diagrams, which illustrate the band bending across the In_2O_3 active layer under applied V_{DS} and V_{GS} biases (Figure 8(b-d)). These biases naturally induce shifts in the quasi-Fermi levels. In constructing the band diagrams, the heavily doped p⁺-Si gate is treated as a metallic electrode, analogous to the Au source and drain contacts. Along the horizontal direction (Au/Ti/ In_2O_3 /Ti/Au stack), the band profile is primarily modulated by V_{DS} , while along the vertical direction (Au/Ti/ In_2O_3 / Al_2O_3 / SiO_2 /p⁺-Si stack), band evolution is dominated by V_{GS} . At $V_{DS} = 0$ V and $V_{GS} = 0$ V, the observed band bending within the In_2O_3 channel, progressing from source to channel center and toward the drain (Figure 8(a)), is consistent with the expected behavior of TFT, irrespective of the presence of a Ti/ In_2O_3 intermixing layer. A key factor influencing these band profiles is the interfacial reaction between Ti and In_2O_3 . Oxidation of Ti, accompanied by reduction of In_2O_3 , redistributes charge at the interface. While partial oxidation of Ti does not significantly alter its metallic character thus maintaining effective charge screening the reduction of n-doped- In_2O_3 , which has a comparatively low carrier density, leads to substantial changes in its conduction properties. As confirmed by ESI, p. S-7, Figure S10, Ti undergoes nearly complete oxidation, while In_2O_3 is strongly reduced, profoundly impacting the interface electronic structure. Additionally, evidence of anomalous charge building-up is observed at the In_2O_3 / Al_2O_3 and Al_2O_3 / SiO_2 junctions, inferred from the non-uniform

global binding energy shifts under applied V_{GS} . At $V_{DS} = 0$ V and $V_{GS} = 100$ V, only minor modifications occur in the horizontal (channel) band profile, whereas significant band bending is induced along the vertical (gate) direction. When both $V_{DS} = 10$ V and $V_{GS} = 100$ V are applied, simultaneous downward band bending is observed along both the horizontal (source-drain) and vertical (source-gate) directions.

To further explore the unexpected global shifts observed in the Al 1s and Si 2p_{3/2} peaks at $V_{DS} = 10$ V (see Figure 5 (b-c)), we additionally performed a position-dependent analysis of these core levels as detailed in the ESI, p. S-5, Figure S7 and p. S-6, Figure S8. The observed global energy shift is corroborated by the behavior of the Au (4f_{7/2}) signal (ESI, p. S-6, Figure S7). At $V_{DS} = 10$ V, the Au 4f_{7/2} peak shifts from approximately 84 eV (source electrode) to around 94 eV (drain electrode). For Al 1s and Si 2p_{3/2}, we note a global shift and position-dependent variation at $V_{DS} = 10$ V and $V_{GS} = 0$ V compared to $V_{DS} = 0$ V and $V_{GS} = 0$ V. The extent of the energy shift is notably larger for the Au 4f_{7/2} (about 10 eV for the Au metal electrode) than for Al 1s and Si 2p_{3/2} (approximately 7 eV for these oxide materials). The position-dependent global shifts observed in the photoelectron signals from Al (Al₂O₃) and Si (SiO₂) are attributed to variations in the electrostatic potential within the In₂O₃ layer, rather than to changes in the chemical bonding of the Al₂O₃ and SiO₂ layers.

As depicted schematically in the ESI, p. S-6, Figure S9, there is an horizontal potential gradient from the drain to the source along the longitudinal axis of the channel supported by the data and also previously reported by Kryvchenkova et al.²⁴. Additionally, there is a constant vertical potential gradient from the Al₂O₃/In₂O₃ interface to the In₂O₃ surface. Consequently, as initial photoelectrons from SiO₂ or Al₂O₃ layers traverse the Al₂O₃/In₂O₃ interface, they experience an additional energy shift related to the horizontal potential electric due to the V_{DS} applied on the drain-source direction of the channel. In other words, at $V_{DS} = 0$ V, the energy shifts observed in the Al 1s and Si 2p_{3/2} peaks primarily reflect a genuine vertical potential variation. In contrast, at $V_{DS} = 10$ V, the measured shifts include an additional contribution arising mostly from the lateral potential gradient of the In₂O₃ semiconducting channel between source and drain, which is mixed or superimposed on the vertical component observed at $V_{DS} = 0$ V. Furthermore, we cannot exclude the possibility of a slight intrinsic modification of the Al 1s and 2p_{3/2} binding energies induced by the applied V_{DS} .

Based on the experimental data from both electrical and chemical analyses, we have studied the impact of these factors on the response of the In₂O₃-TFT under various voltage bias conditions (V_{DS} and V_{GS}). This analysis allows us to propose mechanisms for charge carrier transport of the transistor under operating conditions. Since leakage currents in oxide TFTs such as In₂O₃ may originate from multiple sources (e.g., oxygen vacancies, gatesource leakage, interface traps, heat, bias voltage, percolation), it is not possible at this stage to unambiguously identify the specific and

primary or dominant mechanism responsible for the drain leakage. However, we will describe three distinct stages associated with three different polarities of V_{GS} at $V_{DS} = 10$ V, as shown in Figure 9.

When V_{GS} is at 0 V (Figure 9(a)), naturally the gate dielectric becomes negatively biased relative to the channel, which has a potential gradient along it, with the source grounded and the drain positively biased ($V_{DS} = 10$ V). As a result, any horizontal point on the channel (except at the source terminal) must be more positive than the Al₂O₃ gate dielectric. Thus, the junction formed between the In₂O₃ conducting channel and the Al₂O₃ gate dielectric is reverse-biased, creating a depletion layer that extends into the channel. This reverse bias causes a higher potential across the junction closer to the drain, resulting in a thickening of the depletion layer. Due to the potential gradient along the channel, the shape of the depletion layer is asymmetrical with a maximum in the center of the channel, as depicted in Figure 9(a). The depletion layer is generally thicker towards the drain end of the channel because the voltage on the drain is more positive than the source, creating a voltage gradient along the channel.

With no external V_{GS} and an applied V_{DS} of 10 V, electron charge carrier transport (I_{DSS}) occurs through defects and shallow traps in the channel. Generally, the magnitude of I_{DSS} is limited by the size of the depletion region around the junctions. However, comparing the intensity of I_{DSS} (Figure 3(a)) at $V_{GS} = 0$ V and $V_{GS} = -20$ V with $V_{DS} = 10$ V, we observe that the I_{DSS} values are similar. This observation suggests that in this case the I_{DSS} is not primarily controlled by the depletion region size, which should result in less I_{DSS} at $V_{GS} = -20$ V than at $V_{GS} = 0$ V. The similar I_{DSS} magnitudes can be explained by the intrinsic presence of V_O^{2+} , which induces the creation of localized states within the bandgap of the material and increases the conductivity of oxide materials by providing additional free electrons to the conduction band^{25,26}. Therefore, with $V_{DS} = 10$ V, electrons may hop (or tunnel) between these localized states, contributing to a form of transport known as hopping (tunneling) conduction, particularly in disordered systems like In₂O₃.

As shown in Figure 9, the depletion region at the Al₂O₃/In₂O₃ interface does not extend beneath the Au/Ti electrodes. This spatial confinement to the 130 μ m channel length is supported by observations and analysis of a chemically similar 130 μ m channel In₂O₃-TFT with Au (5nm)/Ti (5nm) source/drain electrodes as presented in the ESI, p. S-7, Figure S10. These thin Au/Ti electrodes allowed us to directly study the Ti/In₂O₃ interfaces beneath the source/drain terminals under different V_{GS} and V_{DS} conditions. The In 3d_{5/2} results indicated a progressive reduction of In₂O₃, leading to the presence of In⁰. We also observed that the relative ratio of In⁰ increases over time and with increasing bias voltage. However, we did not record any binding energy shift of the In 3d_{5/2} as a function of the applied bias voltages or electrode types (source or drain side). From this, we can infer that the depletion layer tends to vanish under the drain/source terminals in the presence of In⁰. This outcome is consistent with the findings of Lee et al.²⁷, who employed TCAD modeling to estimate the channel potential profile. Their results

indicate that the free electron concentration beneath the drain and source electrodes differs from that in the channel, with minor disturbances present at the interface between the electrodes and the SiO_x etch-stop layer. These observations align with our model, which demonstrates that the chemical and electronic states beneath the drain and source electrodes differ from those in the channel. In particular, the reduction of indium oxide beneath the electrodes, induced by the formation of TiO_x , is likely to modify the energy barrier. It should be noted that, in these transistors, the source/channel and channel/drain interfaces are designed to be ohmic (through the formation of a $\text{Ti}/\text{In}_2\text{O}_3$ mixing layer) to facilitate current flow. Figure 9(b) illustrates the ON-state channel conduction along with the residual I_{DSS} . When $V_{GS} > 0$ V is applied, the front side of the Al_2O_3 gate dielectric becomes positively polarized. This polarization causes electron carriers from the In_2O_3 layer to accumulate and be confined at the interface with the gate dielectric, forming a conductive channel path (indicated by yellow arrows). Upon applying $V_{DS} = 10$ V, the accumulated carriers drift from the source to the drain terminals, generating the drain current (I_{DS}).

Based on Figure 3(a), we can argue that I_{DS} at $V_{GS} > 0$ V comprises both the off-leakage current and the intrinsic on-current contribution. Therefore, I_{DS} increases with rising V_{GS} and V_{DS} until reaching saturation, when V_{DS} is sufficiently large to pinch off the channel near the drain side. Moreover, the charge carrier dynamics in an n-channel oxide semiconductor under $V_{GS} > 0$ V and $V_{DS} > 0$ V are complex. In an n-type AOS-TFT, electron transport in the semiconductor channel involves electron hopping between neighboring cations through a percolation-conduction path^{25,28,29}. For effective transport, it is necessary to fill the localized states between the energy bands of the semiconductor before carriers contribute to conduction. Once $V_{GS} > 0$ V is applied, the positive charge on the front side of Al_2O_3 attracts a substantial number of electron carriers to the $\text{In}_2\text{O}_3/\text{Al}_2\text{O}_3$ interface. These carriers first fill the lower-lying localized states of In_2O_3 , and then the remaining electrons fill the upper-lying localized states. With neighboring localized states filled with electrons, remaining carriers can jump to adjacent ions, forming a conduction channel from the source to the drain terminals, as depicted in Figure 9(b).

Finally, when V_{GS} is reversed to negative polarity, as shown in Figure 9(c), the front side of the Al_2O_3 becomes negatively charged, resulting in the formation of a depletion layer with a wider envelope than that observed at $V_{GS} = 0$ V. As previously explained, the primary contribution to the I_{DSS} in our In_2O_3 -TFT arises from electron displacement through V_O^{2+} defects rather than the size of the depletion layer envelope. This observation aligns with the electrical data from Figure 3(a), which shows similar I_{DSS} values at $V_{GS} = 0$ V (small depletion layer) and $V_{GS} = -20$ V (large depletion layer).

IV. CONCLUSIONS

We conducted a non-destructive *operando*-HAXPES analysis on actual In_2O_3 -TFT structures under working conditions.

This investigation elucidates the connection between chemical changes in the channel/gate dielectric structure and the electrical responses under varying V_{GS} and V_{DS} . Our findings from the core level analysis reveal that the binding energy of In 3d shifts along the horizontal direction of the channel layer, influenced by the potential gradient induced by V_{DS} , resulting in differential energy shifts from the drain to the source terminals. Additionally, the polarity of V_{GS} under positive V_{DS} significantly impacts the channel structure, and V_{DS} applied to the channel affects the vertical electrostatic potential of the buried gate dielectric and gate electrode. The primary contribution to the drain-source I_{DSS} is attributed to intrinsic high electron and defect/trap levels due to oxygen vacancies (V_O^{2+}). Furthermore, the investigation of In_2O_3 beneath the Au/Ti drain and source electrodes reveals a significant reduction of In_2O_3 , leading to substantial amounts of In^0 .

SUPPORTING INFORMATION

Optical microscopy images of In_2O_3 -TFTs; Current Voltage recorded during *Operando*-HAXPES analysis at spot II; Current Voltage recorded during *Operando*-HAXPES analysis at spot III; HAXPES spectra of Si 2s and In 4s/Al 2s core level from the center of wide 130 μm length channel; HAXPES spectra of In 3d core level from the center of wide 130 μm length channel; HAXPES spectra of (a) In 3d, (b) Al 1s, (c) Si 2p_{3/2}/Au 4f and (d) O 1s core level from the center of narrow 70 μm length channel; Al 1s core level shifts under varying V_{GS} and V_{DS} from wide 130 μm length channel; Si 2p (Au 4f) core level shifts under varying V_{GS} and V_{DS} from wide 130 μm length channel; Schematic potential distribution and the Al 1s and Si 2p binding energy change through the In_2O_3 semiconducting layer; Deconvolution of the In 3d and Ti 2p core levels from on the Au(5 nm)/Ti(5 nm) thin (a) drain and (b) source electrodes;

DECLARATION OF COMPETING INTEREST

The authors have no conflicts to disclose.

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