Reliable operation in high-mobility indium oxide thin film transistors

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Abstract

Transparent oxide semiconductors (TOSs) based thin-film transistors (TFTs) that exhibit higher field effect mobility (μ_{FE}) are highly required toward the realization of next-generation displays. Among numerous types of TOS-TFTs, In₂O₃-based TFTs are the front-running candidate because they exhibit the highest $\mu_{FE} \sim 100 \text{ cm}^2 \text{ V}^{-1} \text{ s}^{-1}$. However, the device operation of In₂O₃ TFTs is unreliable; a large voltage shift occurs especially when negative gate bias is applied due to adsorption/desorption of gas molecules. Although passivation of the TFTs is used to overcome such instability, previously proposed passivation materials did not improve the reliability. Here, we show that the In₂O₃ TFTs passivated with Y₂O₃ and Er₂O₃ films are highly reliable and do not show threshold voltage shifts when applying gate

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bias. We applied positive and negative gate bias to the In_2O_3 TFTs passivated with various insulating oxides and found that only the In_2O_3 TFTs passivated with Y_2O_3 and Er_2O_3 films did not exhibit threshold voltage shifts. We observed that only the Y_2O_3 grew heteroepitaxially on the In_2O_3 crystal. This would be the origin of the high reliability of the In_2O_3 TFTs passivated with Y_2O_3 and Er_2O_3 films. This finding accelerates the development of next-generation displays using high-mobility In_2O_3 TFTs.

Introduction

Transparent oxide semiconductor (TOS)-based thin-film transistors (TFTs) have been widely used as the backplane of flat panel displays (FPDs), such as liquid crystal displays (LCDs) and organic light-emitting diodes (OLEDs)¹⁻⁴. Their performance is characterized by the field effect mobility (μ_{FE}). The μ_{FE} of amorphous InGaZnO₄ (a-IGZO) TFTs^{5,6}, which has been commercialized, is ~10 cm² V⁻¹ s⁻¹. However, even higher-definition (*e.g.*, super high definition, SHD) FPDs that are operatable at higher frequencies (>240 Hz) demand μ_{FE} ~100 cm² V⁻¹ s⁻¹, one order of magnitude higher than that of a-IGZO TFTs⁷⁻¹⁰. Therefore, numerous efforts have been made thus far to find good TOSs that show high μ_{FE} .¹¹⁻¹⁷ Among numerous reports on TOS-TFTs, Magari *et al.* reported that hydrogen incorporated In₂O₃ TFTs stand out high μ_{FE} ~140 cm² V⁻¹ s^{-1 17}. Although the influence of the grain boundary scattering of electrons in polycrystalline materials is a major concern, the low-temperature (~300 °C) solid-phase crystallization technique is effective in increasing the lateral grain size and mobility and reducing the subgap defects in In₂O₃ films. Thus, In₂O₃ TFTs have been front-running candidates.

However, there is a serious drawback to the operation of In_2O_3 TFTs, which makes them inappropriate for practical applications. In real applications, the TFT operation must be stable against gate bias applications. Current In_2O_3 TFTs exhibit a large negative threshold voltage (V_{th}) shift when a negative gate bias stress is applied (**Fig. 1a**). In the case of a-IGZO TFTs, the origin of the V_{th} shift is mainly gas adsorption/desorption, and such a V_{th} shift is suppressed by covering the active material layer (passivation).^{18,19} As a passivation material, a-SiO₂ is widely used, and the V_{th} shift of a-IGZO TFTs is successfully suppressed.²⁰⁻²⁴ Therefore, Magari *et al.*¹⁷ deposited a-SiO₂ film as the passivation layer of the In₂O₃ TFTs, but the resultant TFTs showed a large V_{th} shift (~5 V) after applying negative gate bias stress. Thus, a-SiO₂ passivation is useless in improving the reliability of In₂O₃ TFTs. To overcome this difficulty and realize ideal In_2O_3 TFTs with reliability (**Fig. 1b**), we passivated the In_2O_3 active layer surface of In_2O_3 TFTs with various insulating oxides, as schematically shown in **Fig. 1c**, including HfO₂, Al₂O₃, Y₂O₃, Er₂O₃, Gd₂O₃, Yb₂O₃, Sm₂O₃, and Nd₂O₃. We selected amorphous HfO₂ and Al₂O₃ films because they are often used as the passivation layers for TOS TFTs.²⁵⁻²⁷ Further, we selected Ln_2O_3 (Ln: Lantanoid, Ln = Y, Er, Gd, Yb, Sm, and Nd) because the ionic radius of the Ln^{3+} is close to that of In^{3+} and the crystal structure of Ln_2O_3 (C-rare earth structure: Ln = Y, Er, Gd, and Yb, B-rare earth structure: Ln = Sm, A-rare earth structure: Ln = Nd) is similar to that of In_2O_3 (C-rare earth structure). Note that Nomura *et al.* reported that Y₂O₃ passivation effectively improves the reliability of a-IGZO TFTs.²⁸ We expected that heteroepitaxial growth of Ln_2O_3 would occur on $In_2O_3^{29}$ and the reliability would be improved. Interestingly, only the In_2O_3 TFTs passivated with Y₂O₃ and Er₂O₃ films showed excellent reliability, whereas those passivated with all other insulating oxides exhibited poor reliability.

Here, we show that In_2O_3 -TFTs passivated with Y_2O_3 and Er_2O_3 films are highly reliable; the V_{th} shifts are negligible after applying a gate bias. We applied positive and negative gate biases to In_2O_3 TFTs passivated with various insulating oxides and found that only the In_2O_3 TFTs passivated with Y_2O_3 and Er_2O_3 films did not exhibit V_{th} shifts. We observed that only Y_2O_3 grew heteroepitaxially on the In_2O_3 crystals. This could be the origin of the high reliability of the In_2O_3 TFTs passivated with Y_2O_3 and Er_2O_3 and Er_2O_3 . The present findings are expected to accelerate the development of next-generation displays using In_2O_3 TFTs showing high μ_{FE} .

Results and discussion

We fabricated bottom-gate top-contact In₂O₃ TFTs on ITO-coated alkali-free glass substrates. The ITO was used as the bottom gate electrode. We used a 100-nm-thick AlO_x film as a gate insulator. The dielectric permittivity (ε_r) and capacitance (C_i) of the AlO_x films were 8 and 70 nF cm⁻², respectively. We deposited a 5-nm-thick amorphous In₂O₃ film on the AlO_x film at room temperature through a stencil mask. The multilayer film was heated at 300 °C in air for 0.5 h to crystalize the In₂O₃ active layer. Then, 100-nm-thick indium tin oxide (ITO) source/drain electrodes were deposited through a stencil mask. The channel length (L) and width (W) were 200 and 400 µm, respectively. After that, ~50-nm-thick HfO₂, Al₂O₃, and Ln_2O_3 (Ln: Lanthanoid = Y, Er, Gd, Yb, Sm, and Nd) films were deposited as the passivation layer through a stencil mask without substrate heating. Finally, the TFT was thermally

annealed at 370 °C in air for 0.5 h. After the thermal annealing, Ln_2O_3 films were polycrystalline whereas HfO₂ and Al₂O₃ films remained amorphous (**Fig. S1**).

The transistor characteristics were measured using a semiconductor device analyzer (B1500A, Agirent) by applying a constant drain voltage (V_d) of 5 V. We measured the transfer characteristics after applying a positive gate bias stress (PBS; $V_g = +20$ V, electric field E = 2 MV cm⁻¹) and a negative gate bias stress (NBS; $V_g = -20$ V, electric field E = -2 MV cm⁻¹) for up to 5000 s to the TFTs at room temperature in air. Details of our TFT fabrication and measurement are described in the **Method** section.

First, we measured the transistor characteristics of the In_2O_3 TFTs without a passivation layer (**Fig. 2**). The TFT showed excellent transistor characteristics before applying bias stress (Holding time = 0 s in **Fig. 2a**); the field effect mobility (μ_{FE}) of ~85 cm² V⁻¹ s⁻¹, threshold voltage (V_{th}) of -3.2 V, subthreshold swing of ~0.22 V decade⁻¹, and TFT did not show a hysteresis behavior. However, after applying PBS for 5000 s, a positive V_{th} shift of +1 V was observed (**Fig. 2a**). After applying NBS for 5000 s, the negative V_{th} shift of -10 V occurred (**Fig. 2b**). These results demonstrate that the polycrystalline In_2O_3 TFTs without a passivation layer are unreliable.

Then, we measured the transistor characteristics of the In₂O₃ TFTs with passivation layers (**Figs. S2–S9**). Although HfO₂ and Al₂O₃ play as good passivation materials for a-IGZO TFTs, the In₂O₃ TFTs passivated with HfO₂ (**Fig. S2**) and Al₂O₃ (**Fig. S3**) show a large V_{th} shift especially after NBS (HfO₂: -4.4 V, Al₂O₃: -7.0 V), showing unreliability of the TFTs. Importantly, the In₂O₃ TFTs passivated with the Y₂O₃ (**Fig. S4**) and Er₂O₃ (**Fig. S5**) films exhibited excellent reliability, and the V_{th} shifts were negligible after both PBS and NBS. The TFTs passivated with the Gd₂O₃ (**Fig. S6**) and Yb₂O₃ (**Fig. S7**) films showed good reliability. However, the TFTs passivated with the Sm₂O₃ (**Fig. S8**) and Nd₂O₃ (**Fig. S9**) films exhibited a large V_{th} shift after both PBS and NBS. The V_{th} shift results after 5000 s application of PBS and NBS are summalized in **Figs. 2c** and **2d**, respectively. From these results, we conclude that the Y₂O₃ and Er₂O₃ films are excellent passivation materials for realizing reliable In₂O₃ TFTs.

We also realized that the performance degradation of the In₂O₃ TFTs did not occur using Y_2O_3 and Er_2O_3 film passivation. Figure 3 shows the bias stress time dependence of the μ_{FE}

(Figs. 3a and 3b) and subthreshold swing (Figs. 3c and 3d) under PBS and NBS. The μ_{FE} and subthreshold swing values of the In₂O₃ TFT without passivation drastically changed with stress time, showing unreliability. In contrast, the μ_{FE} and subthreshold swing values of the In₂O₃ TFTs passivated with Y₂O₃ and Er₂O₃ are highly stable against the bias stress application. The μ_{FE} of the Y₂O₃ passivated TFT was ~78 cm² V⁻¹ s⁻¹ and that of the Er₂O₃ passivated TFT was ~70 cm² V⁻¹ s⁻¹. The subthreshold swing of both TFTs was ~0.1 V decade⁻¹. These results demonstrate that a reliable device operation in In₂O₃ TFTs can be achieved using Y₂O₃ and Er₂O₃ films as passivation layers.

Here, we discuss the origin of the reliability of In₂O₃ TFT passivated with Y₂O₃ and Er₂O₃ films. We hypothesized that the Y_2O_3 and Er_2O_3 films grew heteroepitaxially on the In_2O_3 films because these oxides have the same crystal structure as In₂O₃ (C-rare-earth structure). To confirm this, we prepared ~100-nm-thick rare-earth oxide (Ln_2O_3) (Ln = Y, Nd, and Sm)films on (111)-oriented 10%-Sn-doped In₂O₃ epitaxial films using the same procedure as for passivation layer fabrication. Figure 4 shows cross-sectional high-angle annular dark-field (HAADF) scanning transmission electron microscopy (STEM) images of the Ln₂O₃ films. Although several strong contrasts, indicated by yellow arrows owing to cracks, were observed in the Sm₂O₃ and Nd₂O₃ films, such microstructures were not observed in the Y₂O₃ film. Weak contrasts were observed around the heterointerfaces between Sm_2O_3/In_2O_3 and Nd₂O₃/In₂O₃. Interestingly, the lattice structures of Y₂O₃ are visible together with that of In₂O₃ in the magnified HAADF-STEM images (Figs. 5a and 5b), indicating that heteroepitaxial growth occurred. In contrast, the magnified images of the heterointerfaces of the Sm₂O₃/In₂O₃ (Fig. 5c) and Nd₂O₃/In₂O₃ (Fig. 5d) visualize the existence of an amorphous structure, which is the origin of the contrast around the heterointerfaces. These results reveal that heteroepitaxial growth of the passivation materials is the key to improving the reliability of In₂O₃ TFTs.

To further clarify the origin that heteroepitaxial growth of Y_2O_3 occurred on In_2O_3 , we plotted the relationship between the ionic radius and metal–oxygen bond (*M*–O) length in various Ln_2O_3 (**Fig. 6a**). Ln_2O_3 has three different crystal structures (*i.e.* C-, B-, and A-type) depending on the ionic radius. The C-type structures (Yb₂O₃, Er₂O₃, Y₂O₃, and Gd₂O₃) are cubic bixbyite and have the same crystal structure as In_2O_3 with an ionic radius of 0.80 Å. As shown in light blue in **Fig. 6a**, the *M*–O lengths of Er–O and Y–O are close to In–O. Therefore heteroepitxial growth of Y₂O₃ occurred on In_2O_3 . We also plotted the V_{th} shifts after PBS and NBS of the In₂O₃ TFTs as a function of the difference in *M*–O length from In–O length ($\Delta l_{(M-O)-(In-O)}$) (**Fig. 6b**). The *V*_{th} shifts are significant for both PBS and NBS when the shortest $\Delta l_{(M-O)-(In-O)}$ of the passivation layer exceeds 0.15 Å (Gd₂O₃). In contrast, the *V*_{th} shifts are negligible when the shortest $\Delta l_{(M-O)-(In-O)}$ is below 0.10 Å (Y₂O₃ and Er₂O₃). These results indicate that the bias stress stability is highly dependent on the $\Delta l_{(M-O)-(In-O)}$ of the passivation layer.

Finally, we like to discuss the reason why the heteroepitaxial growth of the passivation material effectively improves the bias stress stability of In_2O_3 TFTs. Generally, gas molecules $(O_2, H_2O, CO_2, etc.)$ are adsorbed on the surface of oxide semiconductors such as IGZO^{18,19}, $In_2O_3^{30}$, and $SnO_2^{31,32}$ and drastically suppress the conductivity of the oxide semiconductors. We performed thermal desorption spectroscopy (TDS) measurements of the In_2O_3 films to determine the absorbed gas molecules (**Fig. S10**). We confirmed that the H₂ and CO desorption are negligible while small desorption of O₂ and large desorption of H₂O occur from In_2O_3 film at relatively low temperatures (~60 °C). These results suggest that the desorption of the OH⁻ ion occurs during the NBS test of the In_2O_3 TFTs and the ion leaves carrier electrons in the In_2O_3 film. Thus, NBS increases the residual carrier electron concentration, resulting in significant V_{th} shifts toward the negative side after NBS. Consequently, the heteroepitaxially grown Y_2O_3 and Er_2O_3 passivation layers would effectively prevent the desorption of OH⁻ ions from the In_2O_3 films. These findings are expected to accelerate the development of next-generation FPDs using In_2O_3 TFTs.

Summary

We demonstrated a key interface engineering strategy to stabilize the device operation of high-mobility In_2O_3 TFTs by applying Y_2O_3 and Er_2O_3 films as a passivation layer. We applied positive and negative gate bias to the In_2O_3 TFTs passivated with various insulating oxides and found that only the In_2O_3 TFTs passivated with Y_2O_3 and Er_2O_3 films did not exhibit degradations in V_{th} , μ_{FE} , and subthreshold swing. We observed that only the Y_2O_3 films grew heteroepitaxially on the In_2O_3 crystals. Consequently, protecting the In_2O_3 backchannel surface with Ln_2O_3 films, featuring similar crystal structure and M–O length as In_2O_3 , prevents the desorption/absorption of gas molecules. This could be the origin of the high reliability of the In_2O_3 TFTs passivated with Y_2O_3 and Er_2O_3 . These findings are expected to accelerate the development of next-generation displays using high-mobility In_2O_3 TFTs.

Methods

Fabrication of In₂O₃ TFTs

Bottom-gate top-contact In₂O₃ TFTs were fabricated on ITO-coated alkali-free glass substrates (EAGLE XG[®], Corning[®]). First, a 100-nm-thick AlO_x gate insulator was deposited on the substrate by atomic layer deposition. The dielectric permittivity (ε_t) and capacitance (C_i) of the AlO_x films were 8 and 70 nF cm⁻², respectively. Subsequently, a 5-nm-thick In₂O₃ film was deposited on the AlO_x film via pulsed laser deposition (PLD) under an oxygen pressure of 3 Pa at room temperature through a stencil mask. The multilayer film was annealed at 300 °C in ambient air for 30 min. Subsequently, 100-nm-thick indium tin oxide (ITO) source/drain electrodes were deposited by PLD through a stencil mask. The channel length (L) and width (W) were 200 and 400 µm, respectively. Subsequently, ~50-nm-thick HfO₂, Al₂O₃, and rare-earth oxide (Ln_2O_3) (Ln: Lanthanoid = Y, Er, Gd, Yb, Sm, and Nd) films were deposited via PLD through a stencil mask without substrate heating. These films served as passivation layers. Finally, the TFT was thermally annealed at 370 °C in air for 30 min.

Reliability tests of the In₂O₃ TFTs

The transfer characteristics of the In₂O₃ TFTs were measured in the dark using a semiconductor device analyzer (B1500A, Agilent). Positive gate bias stress (PBS, $V_g = +20$ V) and negative gate bias stress (NBS, $V_g = -20$ V) tests were conducted for In₂O₃ TFTs for 5000 s at room temperature in air ambient. The field effect mobility (μ_{FE}) was calculated from the linear transfer characteristics at a drain voltage (V_d) of 5 V using equation (1).

$$\mu_{FE} = \frac{g_m}{C_i \frac{W}{L} V_d},\tag{1}$$

where $g_{\rm m}$ is the transconductance, $C_{\rm i}$ is the oxide capacitance of the gate insulator (70 nF cm⁻² for AlO_x), and $V_{\rm d}$ is the drain voltage. The subthreshold swing was extracted from $V_{\rm g}$, which required an increase in the drain current ($I_{\rm d}$) from 1 to 10 pA.

Microstructure analyses of the *Ln*₂O₃ films

To visualize detailed microstructures, including lattice images, single-crystalline In_2O_3 substrates are necessary. First, we prepared atomically flat 10%-Sn-doped In_2O_3 epitaxial films on (111)-oriented YSZ single crystal substrates by PLD. Details of the 10%-Sn-doped In_2O_3 epitaxial film growth have been published elsewhere²⁹. We then deposited Ln_2O_3 (Ln = Y, Sm, and Nd) films on the In_2O_3 epitaxial films by PLD under the same conditions used for

TFT fabrication. Finally, the bilayer films were thermally annealed at 370 °C in air for 30 min. The cross-sectional microstructures of the Ln_2O_3 films were observed using a high-angle annular dark-field (HAADF) scanning transmission electron microscope (STEM; JEM-ARM 200CF, JEOL Co. Ltd.) operated at 200 keV. The incident e-beam direction was $<1\overline{10}>$ In₂O₃.

Thermal desorption spectroscopy of the In₂O₃ films

The desorption gases (H₂, H₂O, CO, and O₂) from the In₂O₃ film were measured by thermal desorption spectroscopy (TDS, TDS1200II, ESCO) with a raising temperature rate of 60 °C min⁻¹. The base pressure of the system was 5×10^{-7} Pa. Si substrates with 100-nm-thick thermally grown SiO₂ were used for the TDS measurements.

Supplementary Information

Supplementary Figs. S1–S10 and Table S1

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Figure 1: Passivation strategy towards the improvement of reliability of high-mobility In₂O₃ TFTs.

Schematic transistor characteristics after applying bias stress of (a) current and (b) ideal In_2O_3 TFTs. Current In_2O_3 TFTs show high-mobility (~100 cm² V⁻¹ s⁻¹). However, there is a major drawback; current In_2O_3 TFTs show large voltage shifts upon applying bias stress. Therefore, ideal In_2O_3 TFTs that show good reliability are in high demand. (c) Passivation strategy. Schematic illustration of a bottom-gate top-contact type In_2O_3 TFT with a surface passivation layer. ~50-nm-thick HfO₂, Al₂O₃, Y₂O₃, Er₂O₃, Gd₂O₃, Yb₂O₃, Sm₂O₃, and Nd₂O₃ films were tested as the surface passivation layer. C-rare earth oxides were mainly chosen because their crystal structures are the same as those of In_2O_3 . HfO₂ and Al₂O₃ films were amorphous, whereas the other oxide films were polycrystalline.



Figure 2: Bias stress test results of the In₂O₃ TFTs with various passivation layers. (a, b) Changes in transfer characteristics of the In₂O₃ TFTs without a passivation layer under (a) PBS (+20 V) and (b) NBS (-20 V). Threshold voltage (V_{th}) shifts of +1 V after 5000 s PBS and -10 V after 5000 s NBS were observed. This is the major drawback of In₂O₃ TFTs. (c, d) V_{th} shifts of the In₂O₃ TFTs with various passivation layers after (c) PBS and (d) NBS for 5000 s extracted from **Figures S1–S8**. The In₂O₃ TFTs passivated with polycrystalline Y₂O₃ and Er₂O₃ films show negligible V_{th} shift under both PBS and NBS. The Gd₂O₃ and Yb₂O₃ film passivation results in a small V_{th} shift after the PBS and NBS. Overall, the In₂O₃ TFTs passivated with C-rare earth oxides show small V_{th} shifts after PBS and NBS for 5000 s. Conversely, In₂O₃ TFTs passivated with HfO₂, Al₂O₃, Sm₂O₃, and Nd₂O₃ films still show large V_{th} shifts after NBS, indicating that these films are not effective as the passivation layers for In₂O₃ TFTs.



Figure 3: Effects of the Y₂O₃ and Er₂O₃ films passivation on the transistor characteristics of the In₂O₃ TFTs.

(a, b) Field effect mobility (μ_{FE}). (c, d) Subthreshold swing. The In₂O₃ TFT without passivation exhibits unreliability; Both μ_{FE} and subthreshold swings change with the stress time. In contrast, the In₂O₃ TFTs passivated with Y₂O₃ and Er₂O₃ are highly reliable; the μ_{FE} and subthreshold swing are stable against the bias stress application. The μ_{FE} of the Y₂O₃ passivated TFT is ~78 cm² V⁻¹ s⁻¹ and that of the Er₂O₃ passivated TFT is ~70 cm² V⁻¹ s⁻¹. The subthreshold swing of both TFTs is ~0.1 V decade⁻¹.



Figure 4: Cross-sectional HAADF-STEM images of the *Ln*₂O₃ films on (111)-oriented In₂O₃ epitaxial films (10%-Sn-doped).

(a) Y_2O_3 (C-rare earth), (b) Sm_2O_3 (B-rare earth), and (c) Nd_2O_3 (A-rare earth). The yellow arrows indicate the cracks. No cracks are observed in the Y_2O_3 film, while many cracks are observed in the Sm_2O_3 and Nd_2O_3 films. The incident e-beam direction is $<1\overline{1}0>In_2O_3$. The scale bars are 50 nm.



Figure 5: Cross-sectional HAADF-STEM images (magnified) of the *Ln*₂O₃ films on (111)-oriented In₂O₃ epitaxial films (10%-Sn-doped).

(a, b) Y_2O_3 (C-rare earth), (c) Sm_2O_3 (B-rare earth), and (d) Nd_2O_3 (A-rare earth). Lattice structures are observed in the case of Y_2O_3 film along with In_2O_3 lattice, indicating that heteroepitaxial growth occurs on the In_2O_3 crystal. In contrast, lattice structure is not observed in the case of Sm_2O_3 and Nd_2O_3 films. Further, thin amorphous layers are observed at the heterointerfaces of Sm_2O_3/In_2O_3 and Nd_2O_3/In_2O_3 . The incident e-beam direction is $<1\overline{10}>$ In_2O_3 . The scale bars are 5 nm.



Figure 6: Possible mechanism of suppression of bias stress shift.

(a) Relationship between the ionic radius and metal–oxygen bond (*M*–O) length in various Ln_2O_3 . The *M*–O length linearly increases with the expanding ionic radius of the metal. The light blue indicates the In–O length range. (b) V_{th} shifts after PBS and NBS of the In₂O₃ TFTs as a function of the difference in *M*–O length from In–O length ($\Delta l_{(M-O)-(In-O)}$). The $\Delta l_{(M-O)-(In-O)}$ of the Ln_2O_3 were extracted from **Fig. 6a**. The V_{th} shifts are significant for both PBS and NBS when the shortest $\Delta l_{(M-O)-(In-O)}$ of the passivation layer exceeds 0.15 Å (Gd₂O₃). In contrast, the V_{th} shifts are negligible when the shortest $\Delta l_{(M-O)-(In-O)}$ is below 0.10 Å (Y₂O₃).

Supplementary Information

Reliable operation in high-mobility indium oxide thin film transistors

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Figure S1: XRD patterns of the passivation layers after thermal annealing at 370 °C. The incident angle of the X-ray is 0.5° . (a) Glass substrate, (b) HfO₂, (c) Al₂O₃, (d) Y₂O₃, (e) Er₂O₃, (f) Gd₂O₃, (g) Yb₂O₃, (h) Sm₂O₃, and (i) Nd₂O₃. The passivation layers (100 nm) were deposited on alkali-free glass substrates and then thermally annealed at 370 °C in air. After the thermal annealing, *Ln*₂O₃ films were polycrystalline whereas HfO₂ and Al₂O₃ films remained amorphous.



Figure S2: Bias stress test results of the In₂O₃ TFT passivated with a HfO₂ film. Changes in transfer characteristics under (a) PBS (+20 V) and (b) NBS (-20 V). Threshold voltage (V_{th}) shifts of +1.2 V after 5000 s PBS and -4.4 V after 5000 s NBS are observed, indicating that the HfO₂ film is not effective as the passivation layer for In₂O₃ TFTs.



Figure S3: Bias stress test results of the In₂O₃ TFT passivated with an Al₂O₃ film. Changes in transfer characteristics under (a) PBS (+20 V) and (b) NBS (-20 V). Threshold voltage (V_{th}) shifts of +0.9 V after 5000 s PBS and -7 V after 5000 s NBS are observed, indicating that the Al₂O₃ film is not effective as the passivation layer for In₂O₃ TFTs.



Figure S4: Bias stress test results of the In₂O₃ TFT passivated with a Y₂O₃ film. Changes in transfer characteristics under (a) PBS (+20 V) and (b) NBS (-20 V). The threshold voltage (V_{th}) shifts after 5000 s of PBS and 5000 s of NBS were negligible, indicating the high reliability of the In₂O₃ TFT passivated with the Y₂O₃ film.



Figure S5: Bias stress test results of the In₂O₃ TFT passivated with an Er₂O₃ film. Changes in transfer characteristics under (a) PBS (+20 V) and (b) NBS (-20 V). The threshold voltage (V_{th}) shifts after 5000 s of PBS and 5000 s of NBS were negligible, indicating the high reliability of the In₂O₃ TFT passivated with the Er₂O₃ film.



Figure S6: Bias stress test results of the In₂O₃ TFT passivated with a Gd₂O₃ film. Changes in transfer characteristics under (a) PBS (+20 V) and (b) NBS (-20 V). The Gd₂O₃ film passivation results in a +0.6 V V_{th} shift after the PBS, whereas it results in a small V_{th} shift (-0.4 V) after the NBS.



Figure S7: Bias stress test results of the In₂O₃ TFT passivated with a Yb₂O₃ film. Changes in transfer characteristics under (a) PBS (+20 V) and (b) NBS (-20 V). The Yb₂O₃ film passivation results in a negligible V_{th} shift after the PBS, whereas it results in a rather large V_{th} shift (-1.9 V) after the NBS.



Figure S8: Bias stress test results of the In₂O₃ TFT passivated with a Sm₂O₃ film. Changes in transfer characteristics under (a) PBS (+20 V) and (b) NBS (-20 V). Threshold voltage (V_{th}) shifts of +1.7 V after 5000 s PBS and -4.7 V after 5000 s NBS are observed, indicating that the Sm₂O₃ film is not effective as the passivation layer for In₂O₃ TFTs.



Figure S9: Bias stress test results of the In₂O₃ TFT passivated with a Nd₂O₃ film. Changes in transfer characteristics under (a) PBS (+20 V) and (b) NBS (-20 V). Threshold voltage (V_{th}) shifts of +4.9 V after 5000 s PBS and -6 V after 5000 s NBS are observed, indicating that the Nd₂O₃ film is not effective as the passivation layer for In₂O₃ TFTs.





(a) m/z = 2 (H₂), (b) m/z = 18 (H₂O), (c) m/z = 28 (CO), and (d) m/z = 32 (O₂). A SiO₂/Si substrate was also tested. Large desorption peaks of H₂O and O₂ are observed around 60 °C and 80 °C, respectively, whereas the desorption of H₂ and CO is negligibly small.

Passivation	Structure	PBS (V)	NBS (V)	Judgment
Without	_	+1.0	-10.0	poor
HfO ₂	Amorphous	+1.2	-4.4	poor
Al ₂ O ₃	Amorphous	+0.02	-7.0	poor
Y ₂ O ₃	Polycrystalline (C- rare earth)	+0.1	-0.1	excellent
Er ₂ O ₃	Polycrystalline (C- rare earth)	+0.1	-0.3	excellent
Gd ₂ O ₃	Polycrystalline (C- rare earth)	+0.6	-0.4	excellent
Yb ₂ O ₃	Polycrystalline (C- rare earth)	+0.1	-1.9	fair
Sm_2O_3	Polycrystalline (B- rare earth)	+1.7	-4.7	poor
Nd ₂ O ₃	Polycrystalline (A- rare earth)	+4.9	-6.0	poor

Table S1: Comparison of reliability of In₂O₃ TFTs with various passivation layers.