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Influence of Grain Boundary Scattering on the Field-Effect Mobility of Solid-Phase Crystallized Hydrogenated Polycrystalline In$_2$O$_3$ (In$_2$O$_3$:H)

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Abstract: Hydrogenated polycrystalline In$_2$O$_3$ (In$_2$O$_3$:H) thin-film transistors (TFTs) fabricated via the low-temperature solid-phase crystallization (SPC) process with a field-effect mobility ($\mu_{FE}$) exceeding 100 cm$^2$ V$^{-1}$ s$^{-1}$ are promising candidates for future electronics applications. In this study, we investigated the effects of the SPC temperature of Ar + O$_2$ + H$_2$-sputtered In$_2$O$_3$:H films on the electron transport properties of In$_2$O$_3$:H TFTs. The In$_2$O$_3$:H TFT with an SPC temperature of 300 °C exhibited the best performance, having the largest $\mu_{FE}$ of 139.2 cm$^2$ V$^{-1}$ s$^{-1}$. In contrast, the $\mu_{FE}$ was slightly degraded with increasing SPC temperature (400 °C and higher). Extended X-ray absorption fine structure analysis revealed that the medium-range ordering in the In$_2$O$_3$:H network was further improved by annealing up to 600 °C, while a large amount of H$_2$O was desorbed from the In$_2$O$_3$:H films at SPC temperatures above 400 °C, resulting in the creation of defects at grain boundaries. The threshold temperature of H$_2$O desorption corresponded well with the carrier transport properties; the $\mu_{FE}$ of the TFTs started to deteriorate at SPC temperatures of 400 °C and higher. Thus, it was suggested that the hydrogen remaining in the film after SPC plays an important role in the passivation of electron traps, especially for grain boundaries, resulting in an enhancement of the $\mu_{FE}$ of In$_2$O$_3$:H TFTs.

Keywords: polycrystalline oxide semiconductors; In$_2$O$_3$:H; solid-phase crystallization; thin-film transistors; high mobility

1. Introduction

Amorphous oxide semiconductors (AOSs), represented by amorphous In–Ga–Zn–O (a-IGZO), have now become standard channel materials in thin-film transistors (TFTs) for active-matrix liquid-crystal displays and active-matrix organic light-emitting diode displays [1–3]. This is because a-IGZO has properties that are superior to hydrogenated amorphous Si (a-Si:H), such as a large field-effect mobility ($\mu_{FE}$) of over 10 cm$^2$ V$^{-1}$ s$^{-1}$, an extremely low leakage current, a low processing temperature (<350 °C), and excellent uniformity [4–12]. Although the $\mu_{FE}$ value of a-IGZO TFTs is more than ten times higher than that of a-Si:H TFTs (<1 cm$^2$ V$^{-1}$ s$^{-1}$), the further improvement of $\mu_{FE}$ values is required to expand their range of applications as alternatives to poly-Si TFTs (50–100 cm$^2$ V$^{-1}$ s$^{-1}$) [13]. The optimization of AOS composition is one approach for improving $\mu_{FE}$; for example, an increase in the In ratio due the considerable spatial spread of the In 5s orbital with a large overlap can provide a facile electron transport path with a low electron effective mass [14,15]. Although various compositions, including In–Sn–Zn–O [16,17], In–W–Zn–O [18,19], Al–In–Sn–Zn–O [20], and In–Ga–Zn–Sn–O [21], have been proposed to enhance the $\mu_{FE}$, the value remains insufficiently high to compete with that of low-temperature polysilicon (LTPS) TFTs [22].
In contrast, the crystallization of OSs is another approach for improving $\mu_{FE}$, because the subgap density of states originating from structural disorder and defects can be suppressed via lattice ordering. Although polycrystalline OSs, including In$_2$O$_3$, ZnO, and SnO$_2$, have been investigated as channel materials in early oxide-based TFTs [23–25], they can easily create oxygen vacancies, leading to degenerate semiconductors. In addition, $\mu_{FE}$ degradation due to grain boundary scattering is a serious issue for polycrystalline OSs as well as poly-Si [26,27]. Polycrystalline In$_2$O$_3$ films have been investigated for use as the transparent conductive oxide (TCO) in solar cells. Koida et al. reported a degenerate hydrogen-doped polycrystalline In$_2$O$_3$ (In$_2$O$_3$:H) film with high electron mobility ($100–130$ cm$^2$ V$^{-1}$ s$^{-1}$) produced by solid-phase crystallization (SPC) [28]. Recently, we reported a $\mu_{FE}$ value of 139.2 cm$^2$ V$^{-1}$ s$^{-1}$ for a TFT obtained using hydrogenated polycrystalline In$_2$O$_3$ (In$_2$O$_3$:H) formed via SPC at 300 °C [29]. The obtained $\mu_{FE}$ value is comparable to the Hall mobility of single-crystalline epitaxial In$_2$O$_3$ films ($\sim 160$ cm$^2$ V$^{-1}$ s$^{-1}$) [30]. The as-deposited amorphous In$_2$O$_3$:H was converted into a polycrystalline film with lateral grain sizes of about 140 nm via SPC [29]. However, the effects of SPC temperature on the electrical and structural properties of In$_2$O$_3$:H films are not yet understood in detail.

In this study, we investigated the effects of SPC temperature on electron transport properties in In$_2$O$_3$:H films and TFTs. Hydrogen intentionally doped during sputtering was found to play an essential role in the passivation of defects, especially for the grain boundaries of the films, resulting in an enhanced $\mu_{FE}$ of the TFTs.

2. Materials and Methods

2.1. Fabrication of In$_2$O$_3$:H TFTs

In$_2$O$_3$:H TFTs were fabricated on a heavily doped p-type Si substrate with a 100 nm thick thermally grown SiO$_2$ layer. The doped p-type Si substrate and SiO$_2$ layer were used as the gate electrode and gate insulator, respectively. The 30 nm thick In$_2$O$_3$:H channels were deposited via pulsed direct current (DC) magnetron sputtering, without substrate heating, from a ceramic In$_2$O$_3$ target using a mixture of Ar, O$_2$, and H$_2$ gases. The gas flow ratios of O$_2$ ($R[O_2] = O_2/(Ar + O_2 + H_2)$) and H$_2$ ($R[H_2] = H_2/(Ar + O_2 + H_2)$) were 1% and 5%, respectively. The deposition pressure and DC power were maintained at 0.6 Pa and 50 W, respectively. The base pressure before gas introduction was below 6 × 10$^{-5}$ Pa. After deposition, the In$_2$O$_3$:H films were annealed in ambient air at a temperature range of 200–600 °C for 1 h. After annealing, a 100 nm thick SiO$_2$ film was deposited via reactive sputtering without substrate heating. Subsequently, Al source/drain electrodes were deposited via sputtering. Finally, the In$_2$O$_3$:H TFTs were annealed at 250 °C in ambient air for 1 h. In$_2$O$_3$:H, SiO$_2$, and Al films were deposited through a shadow mask, and both the channel length and width were 300 μm.

2.2. Characterization Methods

The carrier concentration ($N_v$) and Hall mobility ($\mu_{FE}$) of the In$_2$O$_3$:H films were determined by Hall effect measurements (Accent, HL5500PC) using the van der Pauw geometry at room temperature. The local structural changes of the films were evaluated through extended X-ray absorption fine structure (EXAFS) at the BL01B1 beamline in SPring-8. The In K-edge fluorescence XAFS of the films was measured using a 19-element Ge detector with an incident X-ray angle of 2° with respect to the sample surface. The XAFS data were analyzed using the Demeter software packages [31]. The macroscopic structure of the In$_2$O$_3$:H films was observed using electron backscattering diffraction (EBSD) (EDAX-TSL Hikari High-Speed EBSD Detector). The hydrogen concentration in the films was measured using secondary ion mass spectrometry (SIMS) (ULVAC-PHI, ADEPT-1010) with Cs$^+$ as a primary ion. The chemical bonding states of the constituent elements and hydrogen were evaluated from their thermal effusion using thermal desorption spectroscopy (TDS), which was carried out while varying the stage temperature from 50 to 800 °C at a heating rate of 60 °C min$^{-1}$. Reference films of In$_2$O$_3$ under similar conditions maintaining an $R[H_2]$ of 0% (without hydrogen introduction) were deposited for comparison.
3. Results and Discussion

3.1. Electrical Properties of In$_2$O$_3$:H Films

Figure 1a shows the variations in $N_e$ and $\mu_H$ in the 50 nm thick In$_2$O$_3$:H films as a function of the annealing temperature ($T_{ann}$). The $N_e$ of the as-deposited film ($5.7 \times 10^{17}$ cm$^{-3}$) began to decrease rapidly from $T_{ann} = 200$ °C, where SPC occurred. Then, the In$_2$O$_3$:H exhibited an almost constant $N_e$ of $2 \times 10^{17}$ cm$^{-3}$ (an appropriate value for TFT fabrication) over a $T_{ann}$ range of 300–500 °C. After reaching a $T_{ann}$ of 600 °C, the $N_e$ slightly increased to $1.4 \times 10^{18}$ cm$^{-3}$. The $\mu_H$ of In$_2$O$_3$:H increased to 78.6 cm$^2$ V$^{-1}$ s$^{-1}$ after SPC occurred ($T_{ann} = 200$ °C), whereas the $\mu_H$ of the films decreased to $15$ cm$^2$ V$^{-1}$ s$^{-1}$ at a $T_{ann}$ range of 250–350 °C. As $T_{ann}$ further increased, the $\mu_H$ of In$_2$O$_3$:H began to decrease, resulting in a $\mu_H$ of $0.4$ cm$^2$ V$^{-1}$ s$^{-1}$ at $T_{ann} = 600$ °C.

![Figure 1](image_url)

To understand the carrier transport properties of the In$_2$O$_3$:H films, the relationship between $\mu_H$ and $N_e$ for the films with various $T_{ann}$ values is shown in Figure 1b. The changes in electrical properties could be classified into the following three regions: (I) enhanced $\mu_H$ at $T_{ann} \leq 200$ °C; (II) decreased $\mu_H$ with decreasing $N_e$ at $T_{ann} = 250$–350 °C; and (III) decreased $\mu_H$ with constant or increasing $N_e$ at $T_{ann} \geq 400$ °C. The decrease in the $\mu_H$ of the polycrystalline In$_2$O$_3$:H films with increasing $T_{ann}$ in regions II and III was considered to be due to the effects of grain boundary scattering and intragrain scattering. In general, for degenerate transparent conducting oxide materials, the mobility in the grains is determined by an optical method using the Drude model, because optical mobilities are not affected by grain boundary scattering [32,33]. However, it is difficult to determine the optical mobility of In$_2$O$_3$:H films annealed at $T_{ann} \geq 250$ °C because the free electrons are significantly decreased to the order of $10^{17}$ cm$^{-3}$ and the films become non-degenerate semiconductors [34,35]. Therefore, we evaluated the effects of intragrain scattering by measuring the field-effect mobility of In$_2$O$_3$:H TFTs. This is because when a voltage is applied to the gate, a large number of carriers ($10^{19}$–$10^{20}$ cm$^{-3}$) are accumulated at the In$_2$O$_3$:H/gate insulator interface, and the effects of grain boundary scattering can almost be neglected since electrons tunnel through the narrow width (<1 nm) of the grain barriers at high $N_e$ values [36].

3.2. Electrical Properties of In$_2$O$_3$:H TFTs

Figure 2 shows the typical characteristics of the In$_2$O$_3$:H TFTs annealed at $T_{ann}$ values of 200–600 °C. The $\mu_{FE}$ was calculated from the linear transfer characteristics ($V_{ds} = 0.1$ V) using the following equation:

$$\mu_{FE} = \frac{I_{ds}}{W C_{ox} V_{ds}}$$
where $g_m$ is the transconductance, $C_{ox}$ is the oxide capacitance of the gate insulator, and $V_{ds}$ is the drain voltage. $V_{th}$ was defined by gate voltage ($V_{gs}$) at a drain current ($I_{ds}$) of 1 nA, and $SS$ was extracted from $V_{gs}$, which required an increase in the $I_{ds}$ from 10 to 100 pA. The average values and standard deviations ($σ$) of the characteristics of five TFTs on the same substrate are shown in Supplementary Table S1. The In$_2$O$_3$:H TFT annealed at 200 °C did not exhibit any switching (conductive behavior) because the In$_2$O$_3$:H film was still in a degenerated state ($N_e = 5.7 \times 10^{20}$ cm$^{-3}$). The TFTs annealed at 300 °C exhibited the best performance, with the largest field-effect mobility ($\mu_{FE}$) of 139.2 cm$^2$V$^{-1}$s$^{-1}$ and smallest subthreshold swing ($SS$) of 0.19 Vdec$^{-1}$ with an appropriate threshold voltage ($V_{th}$) of 0.2 V (shown in Figure 2c,e). Although the $\mu_H$ of the In$_2$O$_3$:H films decreased to 14.9 cm$^2$V$^{-1}$s$^{-1}$ with a $N_e$ of $2.0 \times 10^{17}$ cm$^{-3}$ after annealing at 300 °C, as shown in Figure 1, extremely high $\mu_{FE}$ values were obtained from the TFTs at higher gate voltages (Figure 2b). Thus, we concluded that the decrease in $\mu_H$ in In$_2$O$_3$:H TFTs with channels annealed at 300 °C was mainly due to an increase in the potential barrier at the grain boundary caused by a decrease in $N_e$, rather than a decrease in intragrain mobility. In contrast, the TFT characteristics were slightly degraded when $T_{ann} = 600$ °C, i.e., the $\mu_{FE}$ decreased and the $SS$ value increased. Comparing $\mu_H$ and $\mu_{FE}$ after $T_{ann} = 600$ °C, the $\mu_H$ significantly deteriorated to 0.4 cm$^2$V$^{-1}$s$^{-1}$, whereas the $\mu_{FE}$ of the TFTs was maintained at 94.6 cm$^2$V$^{-1}$s$^{-1}$. This result suggests that grain boundary scattering is a dominant factor that limits the $\mu_H$ in films annealed at 400 °C and higher.

Figure 2. (a) Transfer characteristics and (b) $\mu_{FE}$ of the In$_2$O$_3$:H TFTs with channels annealed at various temperatures; $T_{ann}$ dependence of (c) $\mu_{FE}$, (d) $SS$, and (e) $V_{th}$ in the In$_2$O$_3$:H TFTs.

3.3. Structural Properties of In$_2$O$_3$:H Films

To investigate the crystallinity of the films, the effect of $T_{ann}$ on the local structure of the In$_2$O$_3$:H films was evaluated using XAFS. Figure 3 shows Fourier-transformed (FT) EXAFS spectra of the In K-edge for the (a) In$_2$O$_3$ and (b) In$_2$O$_3$:H films as a function of the phase uncorrected interatomic distance. The as-deposited In$_2$O$_3$ film without hydrogen introduction during sputtering exhibited three obvious peaks (Figure 3a), which corresponded to the nearest oxygen (In–O) and the second and third nearest In (In–In
and In–In*). Using the values for the crystalline In$_2$O$_3$ powder standard, the interatomic distance ($R$) and Debye–Waller factor ($σ^2$) for the films were determined. The $k$-range of the EXAFS data used in the analyses was $k = 3–14$ Å$^{-1}$ with a $k$-weight of 3. The fitting carried out in the $R$ space was from $R = 1.0–4.0$ Å for the three-shell model. As shown in Table 1, the $R_{\text{In–O}}$, $R_{\text{In–In}}$, and $R_{\text{In–In*}}$ values of the as-deposited In$_2$O$_3$ film without hydrogen were 2.16, 3.35, and 3.82 Å, respectively, which agreed well with the interatomic distance of the In$_2$O$_3$ bixbyite structure (space group Ia3, number 206) [37–39]. When the In$_2$O$_3$ film was annealed at 300 °C, no noticeable changes in $R$ were observed (Figure 3a), while the peak intensity increased in each spectrum, resulting in a decrease in $σ^2$. This result indicates that thermal annealing improved the structural disorder of the films. By introducing hydrogen during sputtering, the second and third nearest peaks disappeared in the as-deposited In$_2$O$_3$:H film, while the intensity of the first nearest peak decreased (Figure 3b), resulting in an increase in $σ^2_{\text{In–O}}$ to 0.0112. In contrast, after annealing at 200 °C, the intensities of all peaks increased significantly and the intensities of the second and third nearest peaks were higher than those of the In$_2$O$_3$ film annealed at 300 °C. These results indicate that medium-range ordering was lost around In in the initial In$_2$O$_3$:H film, whereas medium-range ordering at distances equal to or longer than the second neighbor significantly improved after annealing at 200 °C. This is in agreement with a previous study using electron backscatter diffraction which confirmed that the amorphous state of the initial In$_2$O$_3$:H film and the grain size of the In$_2$O$_3$:H film were enlarged through SPC [29]. Therefore, the increase in $μ_{\text{FE}}$ after annealing at 200 °C, as shown in Figure 1b (region I), is due to an increase in grain size as well as the improvement of the local structural order of the In$_2$O$_3$:H films. As $T_{\text{ann}}$ increased from 200 to 300 °C, the intensity of the first nearest peak was constant, while the intensities of the second and third nearest peaks increased slightly, resulting in a decrease in $σ^2_{\text{In–In}}$ (0.0056) and $σ^2_{\text{In–In*}}$ (0.0049). This result indicates that the medium-range ordering was improved by annealing at 300 °C, which is in good agreement with the high intragrain mobility obtained via $μ_{\text{FE}}$ for the In$_2$O$_3$:H TFT annealed at 300 °C. As $T_{\text{ann}}$ increased from 300 to 600 °C, the intensities of the second and third nearest peaks further increased, whereas $R$ remained almost constant. Despite the improvement in the crystallinity of the In$_2$O$_3$:H films observed when annealing at 600 °C, the $μ_{\text{FE}}$ of the TFTs slightly decreased to 94.6 cm$^2$ V$^{-1}$ s$^{-1}$, as shown in Figure 2c, and the $μ_{\text{HI}}$ significantly deteriorated to 0.4 cm$^2$ V$^{-1}$ s$^{-1}$. The deterioration of the $μ_{\text{HI}}$ of the In$_2$O$_3$:H films (Figure 1b, Region III) and the $μ_{\text{FE}}$ of the TFTs annealed at ≥400 °C could not be explained by local structural changes in the In$_2$O$_3$:H films. Thus, it was suggested that the decreases in the $μ_{\text{HI}}$ and the $μ_{\text{FE}}$ of the TFTs in region III shown in Figures 1b and 2c were mainly due to the formation of defects at grain boundaries.

![Figure 3](image-url)  
**Figure 3.** FT EXAFS spectra of the In K-edge for the (a) In$_2$O$_3$ and (b) In$_2$O$_3$:H films with various $T_{\text{ann}}$ values.
Table 1. EXAFS fitting results for first, second, and third shells in the In$_2$O$_3$ and In$_2$O$_3$:H films.

| $T_{\text{ann}}$ (°C) | 1st Shell (In–O) | 2nd Shell (In–In) | 3rd Shell (In–In*) |
|------------------------|------------------|-------------------|-------------------|
|                        | $R_{\text{In-O}}$ (Å) | $\sigma^2_{\text{In-O}}$ (Å$^2$) | $R_{\text{In-In}}$ (Å) | $\sigma^2_{\text{In-In}}$ (Å$^2$) | $R_{\text{In-In*}}$ (Å) | $\sigma^2_{\text{In-In*}}$ (Å$^2$) |
| In$_2$O$_3$ as-depo.   | 2.16             | 0.0090            | 3.35             | 0.0065            | 3.82            | 0.0066            |
| In$_2$O$_3$ 300        | 2.16             | 0.0073            | 3.36             | 0.0062            | 3.83            | 0.0063            |
| In$_2$O$_3$:H as-depo. | 2.13             | 0.0112            | -                | -                | -              | -                |
| In$_2$O$_3$:H 200      | 2.16             | 0.0072            | 3.36             | 0.0059            | 3.83            | 0.0051            |
| In$_2$O$_3$:H 300      | 2.17             | 0.0072            | 3.36             | 0.0056            | 3.84            | 0.0049            |
| In$_2$O$_3$:H 600      | 2.17             | 0.0072            | 3.37             | 0.0051            | 3.84            | 0.0046            |

To investigate the origin of the deterioration of the $\mu_H$ of In$_2$O$_3$:H films at $T_{\text{ann}} \geq 400$ °C, we performed EBSD measurements. Figure 4a–e depicts the EBSD images along the normal direction for the In$_2$O$_3$:H films with various $T_{\text{ann}}$ values. The as-deposited amorphous In$_2$O$_3$:H film was converted into a polycrystalline In$_2$O$_3$:H film with grain structure embedded in the amorphous matrix at a $T_{\text{ann}}$ of 150 °C (Figure 4b). At a $T_{\text{ann}}$ of 200 °C (Figure 4c), the film was fully crystallized with a grain size of around 200 nm. After that, no significant difference was observed in the crystal grain size with increasing $T_{\text{ann}}$ values up to 600 °C. The corresponding area fractions of the crystalline phase are shown in Figure 4f. All films showed a maximum area fraction for a grain size of ~200 nm; however, a small proportion of the area fraction with a grain size of ~15 nm was increased in the In$_2$O$_3$:H film annealed at 600 °C, as shown in the insets of Figure 4f. Moreover, these small domains were located in between the large grains, as shown in Figure 4e. The results indicate that when the In$_2$O$_3$:H films were annealed at 400 °C and higher, small domains were created at grain boundaries that served as electron traps, resulting in a decrease in the $\mu_H$ in region III shown in Figure 1b.

Figure 4. EBSD images of the (a) as-deposited, (b) 150 °C, (c) 200 °C, (d) 300 °C, and (e) 600 °C annealed In$_2$O$_3$:H films. (f) Area fraction of each grain size obtained from the In$_2$O$_3$:H films with various $T_{\text{ann}}$ values. The inset shows a magnified view of the small area fraction at a small grain size.
To understand the mechanism of structural deterioration at the grain boundaries of In$_2$O$_3$:H films at $T_{\text{ann}} \geq 400 \degree$C, we performed TDS measurements. Figure 5 shows the TDS spectra of H$_2$, H$_2$O, O$_2$, and In for the In$_2$O$_3$ and In$_2$O$_3$:H films. We first note that H$_2$, O$_2$, and In desorption were negligible for both types of films, while the H$_2$O desorption was high for the In$_2$O$_3$:H film in particular. Large amounts of H$_2$O were desorbed at a stage temperature of 400–800 $\degree$C for the In$_2$O$_3$:H film (Figure 5b). The amount of hydrogen in the film was quantitatively evaluated using SIMS, and it was found that $2.6 \times 10^{21}$ cm$^{-3}$ of hydrogen remained in the film after SPC at 300 $\degree$C, which was one order of magnitude higher than that of the In$_2$O$_3$ film. The H$_2$O desorption temperature (400 $\degree$C) for the In$_2$O$_3$:H film corresponded well to a $T_{\text{ann}}$ of 400 $\degree$C, at which the $\mu_{\text{FE}}$ of the films started to degrade. During annealing at a $T_{\text{ann}}$ of 400 $\degree$C and higher, H or –OH inside grains may migrate to grain boundaries and react with a neighboring H at the boundary, resulting in the generation of H$_2$O molecules. As a consequence, defects are formed at grain boundaries. In other words, the presence of H and/or –OH bonds in the In$_2$O$_3$:H film after SPC plays an important role in the passivation of defects, especially for grain boundaries. In general, the SS value of a TFT is strongly affected by defects near the semiconductor Fermi level [2,12]. We recently reported from hard X-ray photoelectron spectroscopy analysis that intentionally introduced hydrogen is effective in reducing defects near the Fermi level in amorphous IGZO [40–44]. The SS values of the In$_2$O$_3$:H TFTs, shown in Figure 2d, increased at a $T_{\text{ann}}$ of 400 $\degree$C and higher, indicating defect creation. Thus, we believe that the H and/or –OH bonds remained in the films after SPC passivated the defects near the Fermi level, leading to the high $\mu_{\text{FE}}$ and steep SS values of the In$_2$O$_3$:H TFT annealed at 300 $\degree$C. Thus, it is worth noting that grain boundary scattering, which is a serious issue for polycrystalline Si TFTs, may not have a strong influence on the $\mu_{\text{FE}}$ of polycrystalline In$_2$O$_3$:H TFTs.

![Figure 5](https://www.mdpi.com/article/10.3390/nano12172958/s1)

**Figure 5.** (a) H$_2$, (b) H$_2$O, (c) O$_2$, and (d) In desorption spectra from as-deposited In$_2$O$_3$ and In$_2$O$_3$:H films.

4. Conclusions

In summary, we investigated the effects of annealing temperature on the electron transport properties of In$_2$O$_3$:H films and TFTs. The changes in the electrical properties of the In$_2$O$_3$:H films were classified into the three following regions.
(I) When $T_{\text{ann}} = 200^\circ \text{C}$, $\mu_\text{H}$ increased by converting amorphous In$_2$O$_3$:H into a polycrystalline In$_2$O$_3$:H film with an increase in grain size.

(II) When $T_{\text{ann}} = 250$–$350^\circ \text{C}$, $\mu_\text{H}$ decreased with decreasing $N_\text{e}$. However, when $\mu_\text{H}$ exhibited low values, the medium-range ordering in the grains improved. The In$_2$O$_3$:H TFT annealed at 300 $^\circ$C exhibited the best performance, with a $\mu_\text{FE}$ of 139.2 cm$^2$ V$^{-1}$ s$^{-1}$.

(III) When $T_{\text{ann}}$ $\geq$ 400 $^\circ$C, although $\mu_\text{H}$ significantly decreased with constant or increasing $N_\text{e}$, the $\mu_\text{FE}$ of the TFTs was maintained at 94.6 cm$^2$ V$^{-1}$ s$^{-1}$. The medium-range ordering of the In$_2$O$_3$:H network was improved by a $T_{\text{ann}}$ of 600 $^\circ$C, while a large amount of hydrogen was desorbed at 400–800 $^\circ$C, resulting in defect creation at the grain boundaries. Thus, it was suggested that the hydrogen remaining in the film after SPC plays an important role in the passivation of electron traps, especially for the grain boundaries of the In$_2$O$_3$:H films, resulting in an enhancement of the $\mu_\text{FE}$. We believe that the SPC-grown In$_2$O$_3$:H TFTs are promising candidates for use in future electronics applications.

Supplementary Materials: The following supporting information can be downloaded at: https://www.mdpi.com/article/10.3390/nano12172958/s1, Figure S1: Variations of transfer characteristics of the In$_2$O$_3$:H TFTs with channels annealed at various temperatures; Table S1: Summary of the TFT properties.

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