Stress Power Dependent Self-Heating Degradation of Metal-Induced Laterally Crystallized n-Type Polycrystalline Silicon Thin-Film Transistors

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Abstract—Self-heating degradation of n-type metal-induced laterally crystallized polycrystalline silicon thin-film transistors is systematically investigated under various stress powers. A two-stage degradation behavior with turnaround effect at the initial stage is characterized. The initial degradation stage is related to breaking of weak Si–H bonds. The floating-body effect by released hydrogen ions is responsible for the observed back-shift of the transfer curve during the initial stress. On the other hand, the normal degradation stage occurs by breaking of strong Si–Si bonds and trap generation at grain boundaries (GBs) and the gate oxide/channel interface. Our model is supported by observed different activation energies related to two degradation stages and a direct observation of the continuous increase in GB trap density during the normal degradation. Furthermore, during the normal degradation stage, an anomalous continuous field-effect mobility increase along with its V_g dependence shift is first observed. It is clarified that this behavior is not a true channel mobility increase, but a consequence of stress-related trap generation.

Index Terms—Field-effect mobility, metal-induced lateral crystallization, polycrystalline silicon (poly-Si), self-heating (SH) degradation, thin-film transistors (TFTs).

I. INTRODUCTION

LOW-TEMPERATURE polysilicon (LTPS) thin-film transistors (TFTs) have attracted much attention for their potential for system-on-panel applications such as full integrated active matrix displays by liquid crystal or by organic light emitting diodes [1]. Although excimer laser crystallization (ELC) is currently a widely employed LTPS technology [2]–[5], recently, metal-induced lateral crystallization (MILC) has been demonstrated as a promising LTPS technology with some unique advantages over other techniques [6]. Currently, high-performance polycrystalline silicon (poly-Si) TFTs have been fabricated by using various technologies [6], [7]. However, device reliability still remains a limiting factor for poly-Si TFTs toward future application in large-scale integration circuits. It is

particularly true for MILC TFTs since, up to now, there are still rarely reported studies on their device reliability [8], [9].

It has been identified for poly-Si TFTs that hot carrier (HC) stress [4], [5], [9]–[14] and self-heating (SH) stress [4], [7], [9], [14]–[18] are two key mechanisms inducing device degradation. In this paper, time-dependent SH degradation behaviors of n-type MILC TFTs under different stress powers were systematically investigated. Channel temperature rise in stressed devices was simulated by the finite-element method. Two features in device degradation that are different from those previously observed for ELC TFTs, namely, two-stage degradation with a turnaround effect and anomalous field-effect mobility degradation, were observed. A unified model was proposed to understand the observed degradation behaviors.

II. EXPERIMENTS

TFTs used in this study were in conventional self-aligned top-gate structure. First, a 50-nm amorphous Si (a-Si) was deposited on an oxidized silicon wafer by low-pressure chemical vapor deposition (LPCVD). After patterning of a-Si active islands and LPCVD of an 80-nm low-temperature oxide (LTO), a crystallization-inducing window was opened through the LTO, and 5-nm Ni was evaporated by electron beam at room temperature. The wafers were subsequently annealed for metal-induced unilateral crystallization of active islands at 550 °C for 24 h in N_2 ambient. After Ni removal, another extended anneal was done at 550 °C for 24 h. Then, LTO was removed, and another 100-nm LTO layer was deposited as gate oxide, followed by LPCVD of 300-nm poly-Si as gate. After gate patterning, a self-aligned phosphorous implantation with a dose of $4 \times 10^{15} \text{ cm}^{-2}$ was introduced to form source and drain, and subsequently activated at 620 °C for 3 h. Contact holes were then opened before aluminum layer sputtering and patterning. Finally, wafers were sintered in forming gas.

MILC TFTs with gate width/length $W/L = 10/6 \text{ } \mu\text{m}$ are stressed and characterized before and after stress by using Agilent 4156C semiconductor parameter analyzer and Vector MX-1100B prober. From transfer curves measured at $V_{ds} = 0.1 \text{ V}$, device characteristics such as threshold voltage V_{th} , on current I_{on} , field-effect mobility μ_{FE} , and subthreshold slope SS are extracted. I_{on} is defined at a gate voltage $V_g = 12 \text{ V}$. V_{th} is determined by linear extrapolation of I_d to zero. The uncertainty of V_{th} extraction is within 0.2 V. Devices are characterized both in the normal and reverse modes. In the reverse mode, device source and drain electrodes are reversely

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Fig. 1. Typical time evolution of transfer characteristic under SH stress condition for an n-channel MILC TFT, measured at $V_{ds} = 0.1$ V. Lines and symbols are for normal and reverse measurement mode, respectively.

assigned with respect to those in the normal mode. All devices for stress test have similar V_{th} of around 5 V.

SH conditions are set with fixed stress V_g of 30 or 35 V, varied stress V_d from 16 to 22 V, and resulting stress power from ~ 21 to ~ 33 mW. Application of such bias stresses avoids HC generation by impact ionization near the drain, so that only SH-induced degradation effect is investigated. Device degradation after being stressed for varied duration from 1 to 10^5 s is characterized by a percentile change in I_{on} , V_{th} , SS , and μ_{FE} with respect to their initial values, i.e., ΔI_{on} , ΔV_{th} , ΔSS , and $\Delta \mu_{FE}$, respectively.

III. RESULTS AND DISCUSSIONS

A. General Degradation Characteristics

Typical time evolutions of the transfer characteristic of an MILC TFT under SH stress, measured in both forward and reverse modes, are shown in Fig. 1. For low and high power stresses, it is seen that the most dramatic degradation occurs at the pseudosubthreshold region [19], clearly resulting in SS , V_{th} , and I_{on} degradation. I_{on} degradation is reduced at the higher V_g region due to the V_g -induced GB barrier lowering effect [20]–[22]. It is noted that I_{on} degradation symmetrically occurs for forward and reverse measurement modes, implying that such SH stress biases cause uniform degradation along the entire TFT channel. This is reasonable since no saturation occurs in our SH bias stresses. Not surprisingly, more degradation is observed under higher stress power condition. Leakage current increases with stress time, under low and high power stresses, and observed in forward and reverse modes. It implies stress-induced trap state generation near the source and drain junctions.

Percentile degradations of I_{on} and V_{th} against stress time under different stress powers from 21.1 to 33 mW are shown in Fig. 2(a) and (b), respectively. From the dependences of ΔV_{th} and ΔI_{on} and for all stress power, it is seen that there is an initial degradation stage in which device characteristics remain little changed. This initial stage of stress remarkably varies from ~ 1 to $\sim 10^3$ s with decreasing stress power. Significant decrease

Fig. 2. Dependence of (a) on current and (b) threshold voltage degradation on stress time under various stress powers.

Fig. 3. Dependence of (a) $|\Delta I_{on}|$ and (b) $|\Delta V_{th}|$ after initial stage on stress time plotted in logarithmic scale for various stress power conditions. The degradation slopes are 1.0 and 0.95, respectively.

in I_{on} and increase in V_{th} continuously occur only after the initial stage. We then define it as the normal degradation stage.

I_{on} and V_{th} degradation characteristics after the initial stage plotted in logarithmic scale are shown in Fig. 3(a) and (b), respectively. Average degradation slopes are determined to be 1.0 and 0.95 for data range $|\Delta I_{on}| \lesssim 50\%$ and $|\Delta V_{th}| \lesssim 100\%$,

Fig. 4. Subthreshold slope degradation against stress time under various stress power conditions.

Fig. 5. Stress-time-dependent μ_{FE} percentile variation under different stress power conditions.

respectively, which strongly suggests an underlying degradation mechanism of defect generation either by dissolution of Si-H bonds [15] or breaking/distortion of Si-Si bonds [10] at grain boundaries (GBs) and/or the gate oxide/channel interface. After long time stress, the degradation slope of I_{on} and V_{th} is seen to be lowered.

SS degradation for various stress powers is shown in Fig. 4. A power-law dependence is also observed. Degradation slope α for medium and high stress power conditions varies from 0.36 to 0.55, with an average value of 0.44, whereas for the two lowest stress powers, α is found to be 0.85 (21.1 mW) and 0.63 (22.7 mW). Hu *et al.* [11] demonstrated that the degradation slope for defect generation was expected to be between 0.5 (diffusion limited) and 1 (reaction limited). Our observation might indicate a transition of the mechanism from reaction limited to diffusion limited when increasing the stress power, thus enhancing the channel temperature rise. Degradation of SS is a consequence of defect generation along the entire channel upon high current stress under SH [17], depending mainly on deep states at the gate oxide/channel interface and/or GBs [5], [14].

B. Anomalous Field-Effect Mobility Degradation

Different degradation behaviors of μ_{FE} are found under low or high stress power conditions, as shown in Fig. 5. Under low stress power conditions, μ_{FE} remains almost unchanged

Fig. 6. Evolution of μ_{FE} versus V_g dependence upon SH stress for (a) low, (b) medium, and (c) high stress power conditions.

or just slightly decreases after long time stress. Under high stress power conditions, however, μ_{FE} continuously increases with stress time until finally saturating/decreasing after 10^4 s stress. The μ_{FE} increase becomes more evident with increasing stress power. Besides μ_{FE} or maximum transconductance (g_{m_max}) variation, a continuous nearly parallel shift of μ_{FE} versus V_g dependence in the positive direction is observed. The evolution of the μ_{FE} dependence upon SH stress for low, medium, and high stress power conditions is shown in Fig. 6(a)–(c), respectively. The shift is more significant for high power stress conditions, but begins to saturate after 10^4 -s stress synchronizing with the $\Delta\mu_{FE}$ saturation (see Fig. 5). In poly-Si TFTs, μ_{FE} or g_{m_max} degradation was commonly observed in HC degradation [5], [10]–[12]. Previously, there were observations of an initial increase in μ_{FE} or g_{m_max} under HC stress due to the channel shortening effect by HC injection into the gate oxide at the drain end [10], [12], [13]. However, to our knowledge, continuous μ_{FE} or g_{m_max} increase in poly-Si TFTs upon SH stress was not previously reported.

The parallel shift of the μ_{FE} dependence is found to be directly related to the stress-induced device V_{th} degradation. This is clearly demonstrated in Fig. 7, in which the time evolution of μ_{FE} dependence at different stress powers is plotted against $V_g - V_{th}$ instead of V_g itself, where V_{th} degrades with stress time. Hence, stress-induced V_{th} shift is deducted from V_g dependence of μ_{FE} . Strikingly, all such plotted μ_{FE} curves at different stress times almost coincide with each other at their rising and falling stages, and there is no “ V_g dependence” shift anymore. In poly-Si TFTs, the device V_{th} is determined to be the V_g under which sufficient increase in channel mobility rather than channel charge inversion is obtained. [9], [19]. Therefore, this correlation of the μ_{FE} shift with V_{th} degradation can be explained by the stress-induced GB trap generation, which delays the increase in channel carrier mobility, similar to our previous observation for MILC poly-Si TFTs under HC stress [9].

To clarify the nature of the anomalous μ_{FE} increase, effective channel mobility μ_{eff} at different stress times is also extracted

Fig. 7. Evolution of μ_{FE} versus $V_g - V_{th}$ dependence upon SH stress for (a) low, (b) medium, and (c) high stress power conditions, indicating that the μ_{FE} parallel shift is directly related to device V_{th} degradation.

Fig. 8. Gate-to-channel capacitance versus V_g (CV) curve for various stress times measured on a device of $W/L = 220/10 \mu m$. Channel inversion voltage V_t determined by the CV is listed.

through the linear region ON-state drain current by

$$I_d = \frac{W}{L} \mu_{eff} Q_{inv} V_d = \frac{W}{L} \mu_{eff} C_{ox} V_{gt} V_d \quad (1)$$

where $V_{gt} \equiv V_g - V_t$, and V_t is the channel inversion voltage within grains. In (1), it is noted that V_t should also degrade with stress time. A separate CV measurement is carried out on a large reference device ($W/L = 220/10$) with the same fabrication process to determine its variation. The SH stress condition is set to be $V_g = V_d = 25$ V, with stress power of 250 mW. CV curves at different stress times are shown in Fig. 8. V_t is determined from the V_g position at one third of the transition edge from depletion to inversion [19], and is listed in the figure. As demonstrated, the transition edge for channel inversion becomes more gradual with increasing stress time. As a result, V_t is slightly degraded from -1.0 V at initial to -2.0 V after 10^5 -s stress. Since here, V_t is the inversion voltage within grains, it is reasonable to assume the same V_t value for other TFTs with different W/L from the reference device. Furthermore, the stress conditions for the reference TFT are also com-

Fig. 9. Evolution of effective channel mobility μ_{eff} versus V_g dependence under (a) low or (b) high stress power condition.

parable to those applied on the devices investigated. Therefore, it is approximated that V_t is little degraded (within ~ 1 V) for the SH conditions investigated. Given the estimated V_t values obtained from the CV measurement on the reference device, from (1), μ_{eff} of devices stressed under low- or high-power SH conditions is obtained and is plotted against V_g in Fig. 9. With no surprise, similar to that of μ_{FE} , the dependence of μ_{eff} also continuously shifts in the positive direction with stress time. However, μ_{eff} monotonically decreases with stress time and more severely decreases under higher power stress conditions. That is, the observed μ_{FE} increase is purely a transconductance increase and not a carrier mobility increase at all. Previous models attribute g_{m_max} variation to the intragrain carrier mobility degradation caused by the stress-induced degradation of intragrain crystallinity [3], or to the channel shortening effect [10] by carrier trapping. However, both models fail to explain the current observation of stress-induced simultaneous g_{m_max} increase and μ_{eff} decrease.

By definition, μ_{FE} is conventionally calculated from g_{m_max} . Considering the definition of μ_{eff} , one can obtain the following relationship [23]:

$$\mu_{FE} = \mu_{eff} + V_{gt}^* \partial \mu_{eff} / \partial V_g \quad (2)$$

where V_{gt}^* is the V_{gt} value corresponding to g_{m_max} . In (2), it is seen that besides μ_{eff} , a differential term is also included in μ_{FE} . Under low stress power condition, the μ_{eff} decrease can be compensated by the increase in both V_{gt}^* and the differential; therefore, μ_{FE} is a little degraded. Under high stress power conditions, V_{gt}^* increases much more significantly; the increase in the second term in (2) even overwhelms the decrease in μ_{eff} , leading to the observed continuous μ_{FE} increase with stress. Furthermore, when the shift of μ_{FE} versus V_g dependence saturates after 10^4 -s stress, synchronically, μ_{FE} ceases to increase and finally begins to decrease [Fig. 6(c)]. As shown by (2), μ_{FE} is not an independent physical variable in poly-Si TFT. In fact, μ_{FE} characteristics can be mathematically rebuilt from μ_{eff} data based on (2).

According to mobility models in poly-Si TFTs [20]–[24], and neglecting carrier mobility degradation by channel

Fig. 10. $\ln \mu_{\text{eff}}$ plotted against $-1/(V_g - V_t)$ for various stress times under (a) low or (b) high stress power conditions. GB parameter S is extracted from the slope of the plot and listed in the figure.

Fig. 11. GB areal trap density N_t normalized by unstressed value N_{t0} depends on stress time in logarithmic scale for three different stress power conditions. N_t increases with time following a power-law dependence. N_{t0} is estimated $\sim 7 \times 10^{11} \text{ cm}^{-2}$ using channel layer thickness $d = 50 \text{ nm}$.

transverse electrical field [23], [24], μ_{eff} can be expressed as

$$\mu_{\text{eff}} = \mu_0 \exp(-S/V_{\text{gt}}) \quad (3)$$

where S is a parameter that is related to the GB trap established potential barrier [23], [24]. Assuming monoenergetic traps of areal density N_t on GBs, S can be approximated as

$$S = q^3 N_t^2 d / 8 \epsilon_s C_{\text{ox}} k T \quad (4)$$

where ϵ_s is the Si permittivity, and d is the channel layer [21] or the inversion layer thickness [20], [22]. Therefore, by plotting $\ln \mu_{\text{eff}} \sim -1/V_{\text{gt}}$, S of stressed devices at different stress times can be extracted from the slope of the plot, as shown in Fig. 10. For unstressed devices, an apparent straight line is obtained in a wide V_g range. After long time SH stress, although the plot gradually deviates from a straight line, the increase in the slope (or GB barrier parameter S) is evident. As a simple estimation, N_t at different stress times normalized by initial GB areal trap density N_{t0} is calculated according to (4) for low, median, and high power stress conditions, as plotted in Fig. 11. It is

Fig. 12. Stress power dependence of I_{on} and V_{th} degradation after 3000-s stress, with stress V_g set as 30 or 35 V.

Fig. 13. Stress power dependence of (a) SS and (b) μ_{FE} variation after 3000-s stress, with stress V_g set as 30 or 35 V.

found that N_t increases with stress time following a power-law dependence. Obviously, more trap generation occurs for higher stress power conditions, however, with about the same power-law index $\alpha \approx 0.16$ for all three conditions. The increase in N_t in SH conditions is a direct evidence that the defect generation mechanism at GBs is primarily responsible for the observed SH degradation. On the other hand, it is also reasonable to expect that SH should induce defect generation at the gate-oxide/channel interface. In fact, the observation of limited distortion in the CV transition edge shown in Fig. 8 should be directly related to trap generation at the gate-oxide/channel interface, to which the deviation from a straight line in Fig. 10 for long time stressed $\ln \mu_{\text{eff}} \sim -1/V_{\text{gt}}$ plots may also be attributed. However, it is the trap generation at GBs rather than at the interface that is considered as the limiting mechanism in the currently investigated SH degradation.

C. Two-Stage Degradation Characteristics

The stress power dependence of device parameters degradation after 3000-s stress, such as ΔI_{on} , ΔV_{th} , ΔSS , and $\Delta \mu_{\text{FE}}$, is shown in Figs. 12 and 13. Clearly, all device characteristic variations correlate with stress power. Two groups of devices with fixed stress V_g of 30 and 35 V are investigated. However,

TABLE I
THERMAL PARAMETERS OF THE MATERIALS USED IN THE SIMULATION

Material	Density kg/m ³	Specific heat capacitance J/(kg·K)	Thermal conductivity W/(m·K)					
Al	2700	900	237					
polysilicon	2330	716.2	15(un-doped)			45(heavily doped) [25]		
Si [26]	2330	700	27°C	127°C	227°C	327°C		
			156	105	80	64		
SiO ₂ [27]	2200	1000	27.7°C	59.7°C	93.6°C	129.2°C	151.3°C	177.9°C
			1.30	1.35	1.39	1.42	1.47	1.50

no clear difference can be discerned between the two groups. It indicates that the stress-power-decided temperature rise within the TFT channel is a dominant factor in SH degradation [15], [18], whereas stress V_g seems to have no direct impact on SH degradation.

The temperature rise in stressed devices is simulated by the finite-element method using the commercial tool ANSYS. The calculation domain is a whole 4-in. wafer with a single TFT structure situated in the center; however, only the central TFT area needs to be fine meshed. The wafer bottom is fixed at a wafer chuck temperature of 27 °C, whereas the boundary condition for all other surrounding surfaces is natural convection with convective coefficient $h = 7.5 \text{ W}/(\text{m} \cdot \text{K})$. For internal heat transfer mechanism, radiation is neglected since the maximum temperature is only about 200 °C, so that simulation is based on heat conduction. Local heat generation power by Joule heating of the SH stress current $\Delta P_{\text{ch}} = I_{d\text{-stress}}^2 \Delta R_{\text{ch}}$ is applied along the inversion channel (set as 5 nm in thickness), which increases along the channel since the local channel resistance ΔR_{ch} increases from the source to the drain. Temperature dependence of thermal conductivity of related materials [25]–[27] is included, as shown in Table I. Transient thermal analysis shows that upon stress application or removal, the channel temperature rising or falling time to reach a steady state is only about several microseconds, which is far less than the stress duration. The simulated steady-state channel peak temperature depending on stress power is shown in Fig. 14, and in the inset, a typical channel temperature distribution at the gate-oxide/channel interface is shown. An obvious linear dependence of the temperature rise on stress power is obtained. However, data fluctuation also occurs due to variation in power distribution along the channel since devices are stressed under different V_d and V_g biases. The channel temperature distribution is similar for all stress power conditions. Peak temperature appears close to ($\sim 1.5 \mu\text{m}$ away from) the drain edge, as shown in the figure inset. It is noted that the simulated peak temperature is comparable to a recent work by Takechi *et al.* [18].

As has been shown in Fig. 2, in the SH degradation, there is an initial stage where device characteristics remain little changed. Significant degradation occurs only after this stage. In Fig. 15, the transfer characteristic of the device subject to initial stress is compared to the unstressed one. A negative parallel shift in the transfer curve, although quite limited, is

Fig. 14. Simulated channel peak temperature at the gate-oxide/channel interface depending on stress power. The inset is a top view of one half of the channel temperature distribution when the stress power is 27.9 mW.

Fig. 15. Comparison of the unstressed device transfer curve to that after the initial stress. A small negative V_{th} shift is indicated. Inset is I_{on} variation depending on the stress time for the same device.

clearly observed, indicating a small reduction in the flatband voltage at the initial stage of stress. Correspondingly, a slight initial I_{on} increase, which is soon overwhelmed by the normal degradation, is also observed. In the figure inset, depending on the transition or turnaround of I_{on} from the initial stage to the normal degradation stage, a characteristic transition time t^* is defined.

Fig. 16. Dependence of transition time t^* on SH stress power. Inset is an Arrhenius plot of the transition time based on the simulated channel temperature rise.

Fig. 17. Dependence of ΔI_{on} at transition time on SH stress power. Inset is an Arrhenius plot of the I_{on} increase at t^* based on the simulated channel temperature rise.

As shown in Fig. 16, t^* remarkably varies from ~ 1 to $\sim 10^3$ s with decreasing stress power from 33 to 21 mW. Since the respective channel temperature rise is obtained by simulation for each stress power condition, an Arrhenius plot of t^* can be illustrated, as shown by the figure inset. The good linearity of the plot indicates that t^* is controlled by a mechanism that can be characterized by an activation energy E_a , which is determined as ~ 2.2 eV from the slope of the plot. Here, it is worth noting that the channel peak temperature is used to estimate E_a . The implication is that the channel high-temperature region should control the device degradation characteristics. This assumption is reasonable since most SH-induced defect generation either by dissolution of Si–H bonds or by breaking/distortion of Si–Si bonds should favorably occur at the channel peak temperature region. In Fig. 17, this turnaround effect is further characterized by the maximum I_{on} increase during the initial stress, i.e., ΔI_{on} at t^* , which also correlates with stress power. Similarly, its corresponding Arrhenius plot is obtained with good linearity with E_a estimated as ~ 0.47 eV.

Although rarely reported in SH degradation, two-stage degradation has been observed by a few groups in other stress conditions, and various mechanisms have been proposed. In dynamic HC stress, Chang *et al.* [10] reported a V_{th} two-stage degradation, but without V_{th} negative shift, and attributed

the initial degradation stage to breaking of weak Si–H/Si–Si bonds and the second stage to breaking of strong Si–Si bonds. Inoue *et al.* [16] showed that a combination of ion contamination effect and SH stress could lead to a similar two-stage degradation. Kouvatso and Davazoglou [28] observed V_{th} turnaround on solid-phase crystallization poly-Si TFTs stressed by high V_g bias with/without a low V_d bias and explained the initial V_{th} backshift by hole trapping of preexisting oxide and interface traps via a mechanism of electron tunneling out of neutral oxide traps [29]. In HC degradation, Farmakis *et al.* [12] proposed that hot hole injection followed by hot electron injection can account for such V_{th} turnaround effect.

In HC two-stage degradation, the initial V_{th} backshift is accompanied by a g_{m_max} increase as a result of the channel shortening effect [10], [12]. In our observation, the lack of initial g_{m_max} increase implies a different underlying mechanism. Based on a previous hypothesis [10], we tentatively propose that at the initial stage of stress, dissolution of Si–H and Si–OH weak bonds at GBs and the gate-oxide/channel interface first occurs by high current stress with elevated channel temperature under SH. Hence, a small amount of released positive hydrogen ions can diffuse toward the bottom oxide under the influence of the transverse electrical field within the channel, slightly modulating the substrate potential. This floating-body effect [7] could be responsible for the initial small negative shift of V_{th} . In the normal stage of stress, degradation is mainly attributed to defect generation by distorting/breaking of strong Si–Si bonds at GBs, resulting in a large increase in V_{th} , which overwhelms the initial V_{th} backshift.

This explanation is supported by our observation that occurrence of the normal degradation stage (t^*) is controlled by a high E_a of 2.2 eV, whereas the amount of initial degradation (ΔI_{on} at t^*) is characterized by a much lower E_a of 0.47 eV. Since E_a extraction is performed by using simulated channel peak temperatures, it is considered that an estimated E_a of 2.2 eV is fairly close to the Si–Si bond energy of 1.83 eV. It lends further support to the hypothesis that the normal degradation occurs by stress-induced defect generation by breaking/distorting of strong Si–Si bonds. Therefore, the carrier trapping mechanism [12], [28], [29] previously proposed to explain the turnaround degradation behavior seems to be not a controlling mechanism in the current observation.

IV. CONCLUSION

SH degradation of n-type MILC polysilicon TFTs is systematically investigated under different stress powers. All degradation characteristics are found to correlate with stress power, indicating the dominant effect of the channel temperature rise in SH degradation. Degradation slopes of device V_{th} , I_{on} , and SS suggest a degradation mechanism of defect generation either at GBs or at the gate oxide/channel interface.

A two-stage degradation behavior with turnaround effect at the initial stage is observed. Normal degradation is caused by breaking/distorting of strong Si–Si bonds mainly at GBs. The turnaround effect at the initial stage is explained by breaking of weak bonds and releasing positive hydrogen ions, which migrate toward the bottom oxide and modulate substrate

potential. Based on the simulated channel temperature, E_a characterizing both degradation stages is estimated. E_a of the initial degradation is found to be much lower than that of the normal degradation.

An anomalous continuous μ_{FE} increase along with its V_g dependence shift is first observed. The V_g dependence shift correlates with device V_{th} degradation. However, μ_{eff} continuously decreases with stress. The anomalous μ_{FE} increase is not a true mobility increase, but is a result of the stress-induced V_g dependence shift in μ_{eff} and the variation of its differential on V_g . Based on a mobility model in poly-Si TFT, where carrier transport is controlled by V_g modulated GB potential barrier, stress-power-dependent GB trap generation is directly observed. All observations are consistent within the same degradation model.

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