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Pulsed Laser Annealing for the Formation of Ultra-Shallow Junctions for Nano-Scale Metal-Oxide-Semiconductor Technologies

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ABSTRACT

The continuous scaling of gate dimensions has created a strong need for better ultra-shallow junction technology to suppress short channel effect, which is more serious for shorter devices. As ion implantation remains the most popular method for dopant introduction, tremendous effort has been drawn into the development of post-implantation annealing techniques. Laser annealing (LA) has been shown to be a promising technique in the formation of ultra-shallow junction due to a number of its favorable features. LA has a near-zero thermal budget that is capable of producing a highly activated and abrupt ultra-shallow junction.

In this work, various aspects of shallow- and non-melt LA were explored. The topics include dopant activation, dopant retention and surface morphology induced by shallow- and non-melt LA. Thermal stability of laser activated boron against post-LA rapid thermal annealing (RTA) was also studied. Thermal confinement effects in advanced substrates and optical coupling effects of Ni capped layer were also studied in this work.

Despite that shallow-melt LA produces a highly activated abrupt junction, a melt-depth related dopant loss could compromise its advantages. Non-melt LA produces a moderately activated junction with negligible diffusion. Dopant activation can be further enhanced using a Si^+ subamorphization implantation prior to non-melt LA. The enhancement is caused by a vacancy-rich surface region generated by the Si^+ pre-implantation that promotes the incorporation of boron atoms into the substitutional sites.

Dopant activation induced by LA exceeds the equilibrium solid solubility of dopant in the Si substrate. The activated dopant exists in a metastable state and can be

deactivated upon subsequent thermal cycles. It is found that laser activated boron atoms remain stable up to 800 °C, after which substantial deactivation takes place in Si⁺ and Ge⁺ preamorphized samples. This is due to the presence of excess interstitials in the end-of-range region that could not be annealed out by melting up to the amorphous/crystalline interface. This deactivation can be significantly suppressed when the melt depth is extended beyond the amorphous layer. A post-LA low temperature RTA of 600 °C has been shown to be critical for the further elimination of the crystalline defects in the laser annealed preamorphized pMOSFET. Good pMOSFET characteristics can be obtained using a combination of LA and a post-LA low temperature RTA, which is suitable for low thermal budget Si process.

Advanced semiconductor substrates, such as silicon-on-insulator and strained-silicon/SiGe, comprise a thermally less conductive layer, which deprives a good thermal dissipation pathway during LA. This gives rise to a highly non-equilibrium laser process, that causes enhanced heating or melting to the substrates than that in normal bulk Si substrate. This thermal confinement effect facilitates non-melt LA in producing a high dopant activation in these advanced semiconductor.

Simultaneous formation of Ni-silicided p^+/n ultra-shallow junction can be achieved using LA on a Ni capped boron implanted Si. The enhanced optical coupling of the capped Ni layer is capable of generating a substantially high temperature, leading to a simultaneous dopant activation and Ni silicide formation. Although the junctions lie near beneath the Ni-Si (or NiSi₂) layer, they show good junction leakage and transistor characteristics.

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LIST OF ABBREVIATIONS

α -Si	amorphous silicon
a/c interface	amorphous/crystalline interface
AFM	atomic force microscopy
BCA	binary collision approximation
BED	boron enhanced diffusion
BIC	boron-interstitial cluster
BOX	buried silicon dioxide
CMOS	complementary metal-oxide-semiconductor (field effect transistor)
c-Si	crystalline silicon
EM	electron multiplier
EOR	end of range
FUSI	fully silicided (gate)
FWHM	full width half maximum
HF	hydrofluoric acid
ITRS	International Technology Roadmap for Semiconductor
I-V	current-voltage (with suffix) or interstitial-vacancy
l/s interface	liquid/solid interface
LA	laser annealing
LPE	liquid-phase epitaxial
LSS	Lindhard, Scharff, Schiott
MOSFET	metal-oxide-semiconductor field effect transistor

nMOSFET	n-channel metal-oxide-semiconductor field effect transistor
o/s interface	oxide/silicon interface
OED	oxygen enhanced diffusion
PAI	preamorphization implantation
PAI-Si	preamorphized silicon
PECVD	plasma enhanced chemical vapor deposition
PIII	plasma immersion ion implantation
pMOSFET	p-channel metal-oxide-semiconductor field effect transistor
RCA	Radio Corporation of America
RMS	root mean square
R_p	projected range
ΔR_p	projected straggle
R_{acc}	accumulation resistance
R_{co}	contact resistance (contributed by silicide)
R_s	sheet resistance (contributed by the doped region)
R_{sp}	spreading resistance
RTA	rapid thermal annealing
S/D	source/drain
SAI	subamorphization implantation
SCE	short channel effect
SIMS	secondary ion mass spectrometry
SLIM	simulation of laser interaction with materials
SOI	silicon-on-insulator

SPE	solid-phase epitaxial
TED	transient enhanced diffusion
TEM	transmission electron microscopy
TESP	tapping mode etched silicon probe
TRIM	transport of ions in matter
ULE	ultra-low energy
x_j	junction depth
x_m	melt depth
XRD	X-ray diffraction
XTEM	cross-sectional transmission electron microscopy

CHAPTER 1

INTRODUCTION

1.1 Background

The success of semiconductor industry over the past few decades has been driven by the continuous scaling down of gate dimension of metal-oxide-semiconductor field effect transistor (MOSFET). Scaling down the gate dimension improves drive current and circuit density, creating faster devices with more functionalities [1],[2]. However, the aggressive scaling down of gate dimension of nano-scale MOSFET has aggravated the short channel effects (SCE), leading to an excessive leakage current to the devices [2],[3]. As a result, the need for ultra-shallow source/drain extensions, as shown in figure 1.1, becomes more critical to suppress the SCE [4]–[7]. The junction depth (x_j), defined as the position where the dopant concentration equates the background concentration of the substrate, of the source/drain extensions must be scaled in conjunction with the scaling of critical dimensions. This requires a stringent control over both lateral and vertical diffusion of the junctions in nano-scale devices. Junctions must be shallow in the lateral dimension for process reproducibility, threshold voltage and drain current control. The vertical dimension must also be shallow to confine the electric field profile in the channel near the gate [2].

The main challenge in ultra-shallow junctions formation lies in achieving both good short-channel characteristics and high current drive. Tradeoff between x_j and sheet resistance (R_s) in the source/drain regions exists, in which R_s increases when x_j decreases without any increase in the active dopant concentration or mobility [8]–[10].

Figure 1.1. Schematic of a MOSFET showing the components of the total resistance.

Optimization of short-channel characteristics and current drive requires strict control in x_j , doping profile and degree of dopant activation. As shown in figure 1.1, the total series resistance in S/D junctions is the sum of contact resistance contributed by silicide (R_{co}), sheet resistance of the doped region (R_s), spreading resistance where carriers turn towards the channel (R_{sp}), and voltage dependent accumulation resistance where the gate overlaps with the junction (R_{acc}). A reduction in the total series resistance would result in increase in current drive [2].

Ultra-shallow p^+/n junctions using boron implantation are more difficult to fabricate compared to n^+/p junctions due to channeling of boron during implantation and transient enhanced diffusion (TED) of boron during post-implantation annealing [8],[9],[11]–[13]. TED arises from the excess interstitials generated during ion implantation and causes boron diffusivity to be several orders larger than that of equilibrium diffusivity. Reducing the thermal budget, the product of temperature and time, of the post-implantation annealing reduces the interaction time of boron atoms with excess interstitials and suppresses TED [8],[13]. It has also been shown that TED is significantly suppressed by reducing the implantation energy to sub-keV regime [13],[14]. The suppression is

attributed to increase interstitial annihilation at the surface. It has been also demonstrated that TED can be minimized by performing a subamorphization Si implantation [15]. The subamorphization implantation reduces channeling of boron and provides vacancies to recombine with the interstitials generated during boron implantation, leading to the reduction of excess interstitials that can contribute to TED.

Figure 1.2. The 2004 ITRS junction depth (x_j) and sheet resistance (R_s) requirements for p-type source/drain extensions [4].

However, diffusivity enhancement of a factor of four still exists where the atomic concentration of boron is greater than 6%. This enhancement, termed boron enhanced diffusion (BED), arises from the formation of Si boride phase (SiB_4) that leads to interstitial injection [16]. Oxygen enhanced diffusion (OED), on the other hand, occurs in the presence of oxygen in annealing ambience. The presence of oxygen leads to oxide growth that injects interstitials into the silicon substrate, resulting in an increased boron

diffusion [17],[18]. Therefore, substantial efforts have been made in the suppression of these anomalous enhanced diffusions.

The 2004 International Technology Roadmap for Semiconductors (ITRS) reveals that source/drain extensions of 14 nm with sheet resistance less than $884 \Omega/\square$ for 35 nm gate length devices are needed [4], as shown in figure 1.2. Junctions of this shallow approach the process limits of conventional ion implantation and thermal annealing processes. Consequently, tremendous efforts have been streamed for the development of new processes for fabrication of highly activated ultra-shallow junctions. These processes include ultra-low energy (ULE) implantation, plasma immersion ion implantation (PIII), spike rapid thermal annealing (RTA), flash annealing, solid-phase epitaxial (SPE) regrowth annealing and laser annealing (LA) [2],[19]–[24] .

Among the developed post-implantation annealing techniques, LA stands out to be the most promising due to its capability of producing highly activated ultra-shallow abrupt junctions. Since each laser pulse lasts only for tens of nanoseconds, LA provides a near-zero thermal budget annealing that raises the substrate surface temperature to greater than 1000°C within tens of nanoseconds [22]–[24]. This minimizes dopant diffusion due to its extremely rapid heating and quenching rate. Typical shallow-melt LA involves a melting of the surface region of the implanted substrate and causes the dopant to be distributed uniformly, producing a box-like profile after a liquid-phase epitaxial (LPE) recrystallization of the melted region [25]. This is possible due to the fact that the diffusivity of boron in liquid silicon is about eight orders higher than that in solid silicon [24]. The highly non-equilibrium LPE recrystallization process also results in a high concentration of dopant activation that exceeds the solid solubility limit due to a solute

trapping mechanism [24]. These features have led to the formation of highly activated ultra-shallow abrupt junctions that are determined by the laser-induced melt depth. Alternatively, LA can be conducted in the non-melt regime, in which the laser irradiation only heats up the substrate to a temperature below the melting temperature of silicon [26],[27]. As a result, the near zero thermal budget of non-melt LA creates a diffusionless ultra-shallow junction defined by the ion implantation.

1.2 Objectives

1. To study the dopant activation and retention in crystalline and preamorphized silicon upon single-pulse and multiple-pulse LA.
2. To explore the dopant activation and the junction leakage characteristics in a Si^+ subamorphized silicon upon LA.
3. To study the effect of substrate structures on dopant distribution after LA and compare the dopant profiles with those obtained in typical bulk silicon substrate.
4. To study the effects of post-LA rapid thermal annealing on the boron deactivation and diffusivity.
5. To study the effect of laser-induced salicidation on dopant activation and diffusion, where the boron implanted samples are capped with silicide metal upon LA.

1.3 Scope of the Project

This project involves the formation and characterization of ultra-shallow p^+/n junctions using LA. For the formation of ultra-shallow junctions, shallow-melt and non-melt LA on crystalline-silicon, Si^+ and Ge^+ preamorphized silicon were explored. The

project also studies dopant activation and the junction leakage characteristics in a Si^+ subamorphized silicon upon LA. The project also delves into the effect of residual defects on boron deactivation and diffusivity enhancement during post-LA rapid thermal annealing. Comparison was made between LA-formed junction profiles in bulk silicon substrate and advanced substrates, such as silicon-on-insulator (SOI) and SiGe substrates. Simultaneous formation of ultra-shallow junctions and nickel silicide using LA are also included in this project. The junctions were mainly characterized using secondary ion mass spectrometry (SIMS), transmission electron microscopy (TEM), sheet resistance measurement and Hall measurement. Other characterization techniques used in this project include atomic force microscopy (AFM), X-ray diffraction (XRD) and Raman spectroscopy. Electrical characterization was also conducted on p^+/n diodes and pMOSFETs formed by LA.

1.4 Organization of the Thesis

The organization and outline of various chapters of this thesis are provided as follows.

Chapter 1 provides the introductory overview pertaining to the formation of ultra-shallow junctions. It also describes the objectives and scope of this project.

Chapter 2 gives a detailed review on the key findings in the earlier work on junction engineering. It covers the background and relevant theories for the formation of ultra-shallow junctions. Fundamentals of LA are also included in this chapter.

Chapter 3 describes the experimental methodology that includes experimental setup, sample preparation, materials and electrical characterization techniques used in this project.

Chapter 4 presents the dopant activation, dopant retention and surface morphology induced by shallow- and non-melt LA. The enhanced dopant activation and the junction leakage characteristics in a Si^+ subamorphized silicon upon LA were also studied.

Chapter 5 studies the effects of post-LA thermal annealing on dopant deactivation and diffusivity enhancement in crystalline-silicon, Si^+ and Ge^+ preamorphized silicon samples. This chapter also reports the advantage of a post-LA low temperature RTA in the further elimination of crystalline defects in the laser annealed Si^+ preamorphized pMOSFET.

Chapter 6 discusses the LA induced thermal confinement effect in advanced semiconductor substrates, such as silicon-on-insulator and strained-silicon/SiGe substrates, that consist of a less thermally conductive structure. Feasibility of the application of non-melt LA on these substrates is also reported.

Chapter 7 reports the simultaneous formation of nickel silicided p^+/n ultra-shallow junction upon LA. The electrical characteristics of the p^+/n diodes and pMOSFET formed with this method are also examined.

Chapter 8 summarizes the major results and findings in this project. It also offers future research direction based on the results of this project.

CHAPTER 2

LITERATURE REVIEW

2.1 Introduction

Junction formation involves a series of delicate processes including dopant introduction, activation, diffusion and defect annihilation. Precise control over dopant dose and position in the substrate can be achieved using ion implantation. Ion implantation, however, places most of the dopant atoms in the interstitial sites and generates a high concentration of lattice defects that lead to undesired leakage current to the junctions. A post-implantation annealing is therefore necessary to activate the dopant by placing the dopant atoms into the substitutional sites and to heal the implantation induced defects. Upon annealing, dopant atoms are activated when they move into substitutional sites but, at the same time, they diffuse further into the substrate bulk leading to a deeper dopant profile. This chapter provides the background and relevant theories on the formation of ultra-shallow junctions. It also covers the relevant theories on the interaction between laser irradiation and materials.

2.2 Ion Implantation

2.2.1 Ion implantation for ultra-shallow junctions

Ion implantation is a process that introduces energetic ions into the substrate with precise control over the number and position of implanted dopant atoms in the substrate [2],[28]-[30]. The process involves an extraction and an acceleration of positively charged

dopant atoms or molecules by a predetermined potential. The process continues with a mass and energy separation through a sector mass analyzer magnet and a series of slits, after which the mono-energetic and mono-species beam is directed into the substrate. The energetic ions travel in the substrate and come to a rest through a series of nuclear collisions, $S_n(E)$, and electronic stopping, $S_e(E)$. Collision cascade upon nuclear collisions can cause the displacement of substrate atoms from lattice locations, leading to a high concentration of Frenkel pair (interstitials and vacancies) point defects, small interstitial and vacancy clusters, point defect-dopant atoms complexes, and amorphous pockets [30]. In electronic stopping mechanism, implanted ion energy is transferred to the electrons leading to the excitation of electrons or ionization of the substrate atoms by extraction of electrons. The total rate of energy loss, dE/dx , in the substrate is given by the sum of these stopping mechanisms [2],[28],

$$\frac{dE}{dx} = S_n(E) + S_e(E) \quad (\text{Equation 2.1})$$

The stopping mechanisms determine the statistical distribution of the implanted dopant atoms in the substrate. The range-energy relation provided by equation 2.1 was formulated by Lindhard, Scharff, and Schiott (LSS) for implantation into an amorphous substrate [28],[31]. The Gaussian implanted dopant profile predicted by LSS is not entirely correct but serves as a useful approximation.

A direct way of creating a shallower implanted dopant profile is through the lowering of implantation energy [2],[5],[13]. A lower implantation energy creates both a smaller statistical projected range (R_p) and projected straggle (ΔR_p), resulting in a shallower as-implanted dopant profile. The challenges in the lowering the implantation energy lie in maintaining a reasonably high current for high throughputs while minimizing energy

contamination caused by ion neutralization [2]. Typical ultra-low energy implantation circumvents the problem by extracting a high beam current at a higher energy and then decelerating the beam to a final low energy [32].

One of the complications in ion implantation is dopant channeling in a crystalline silicon (c-Si) substrate, leading to a substantially long tail in the implanted dopant profile [2],[13],[28]. When the energetic ions are directed into the c-Si substrate, the ions can travel a considerable distance with little energy loss in a certain crystal orientation where obstruction is minimum. The nuclear stopping is not effective and the electron density in this in-between crystal planes or axial tunnels is low. The ions will continue to travel in that direction, making many internal collisions that are nearly elastic until the ions come to rest or de-channel. The channeling effect is more pronounced for light dopant species, such as boron and phosphorus. The channeling effect can be suppressed by titling the (100) c-Si substrate surface by an angle of 7° to prevent the exposure of crystal channels. However, the effect of tilt angle is almost negligible for ultra-low energy ion implantation of light dopant species [13].

The channeling effect can also be reduced by preamorphizing the crystal lattice prior to the dopant implantation [2],[33],[34]. The technique uses a preamorphizing implantation (PAI), such as Si^+ or Ge^+ , to create a continuous amorphous silicon (α -Si) layer from the substrate surface. The α -Si layer de-channels the dopant ions, suppressing the formation of the long dopant profile tail. The PAI scheme, however, generates a supersaturated excess interstitials beyond the amorphous/crystalline (a/c) interface [33]–[35]. The end-of-range (EOR) damage can cause enhanced diffusion [37]–[39] and/or excessive junction leakage current [36] upon a solid-phase epitaxial (SPE) regrowth of the α -Si layer induced by the

post-implantation annealing [37]–[40]. The a/c interface, therefore, has to be far from the depletion region of the junction to prevent undesirable junction leakage. Due a greater atomic mass, Ge^+ PAI produces a sharper a/c interface that inhibits the formation hairpin dislocation during the annealing process [36]. This also reduces the junction leakage current due to the reduction in the formation of extended defects.

Alternatively, a thin screen silicon dioxide can be used to randomize and reduce the ion velocities before entering the substrate [29],[41]. This suppresses the channeling effect and, at the same time, produces an as-implanted profile close to the substrate surface. This technique, however, suffers a significant dose loss due to the trapping of dopant atoms in the oxide layer. Ion implantation through the oxide may also introduce oxygen atoms into the substrate by knocking out the oxygen atoms from the oxide layer.

BF_2^+ implantation produces a shallow B^+ dopant profile as the B^+ ion acquires only ~22% (i.e. the atomic mass ratio of ^{11}B to $^{11}\text{B}^{19}\text{F}_2$) of the implantation energy of BF_2^+ [2],[29]. As this technique produces shallower B^+ dopant profile at a higher implantation energy, it somewhat relieves the stringent technological requirements for ultra-low energy implantation. Moreover, the heavy BF_2^+ ions create an α -Si layer to the surface of the Si substrate during implantation, leading to a suppression of channeling tail. It has been shown that fluorine reduces transient enhanced diffusion (TED) during post-implantation annealing by trapping of Si interstitials [37]. However, BF_2^+ implantation has several disadvantages. Fluorine retards boron activation during post-implantation annealing due to the formation of B-F complexes. The presence of fluorine in to the gate enhances boron penetration into the channel region, leading to a threshold voltage instability [42],[43]. Utilizing the same concept as BF_2^+ implantation, $\text{B}_{10}\text{H}_{14}$ implantation produces shallow B^+

dopant profile by acquiring 11/124 of $B_{10}H_{14}$ implantation energy [23],[44]. $B_{10}H_{14}$ implantation, on the other hand, does not incorporate fluorine atoms that show deleterious effect to the gate oxide. However, $B_{10}H_{14}$ implantation suffers a low beam current and throughput due to a molecule dissociation of $B_{10}H_{14}$ [2],[23],[44].

2.2.2 Plasma doping

Plasma doping or plasma immersion ion implantation is different from the conventional ion implantation because the technique does not use the implanter accelerating column, mass analyzing magnets and ion beam optics [2],[45]–[49]. The substrate is placed in a process chamber surrounded by a plasma ion source and the implantation occurs with a negative bias applied to the substrate. The implantation energy is directly controlled by the applied bias, which can be a very low energy for ultra-shallow junction application. High beam current is possible due to a full wafer implantation in the surrounding plasma. This enables an ultra-low energy implantation at a high beam current, leading to high doping throughput. The technique, however, suffers an energy contamination and dose error due to collisional charge exchange [2]. This non-mass selective implantation process can introduce a great extent of energy and dopant contamination into the substrate.

2.2.3 Infusion doping

Infusion doping uses a gas-cluster ion beam source, producing energetic clusters of dopant atoms [50]–[53]. The gas-cluster ions are then accelerated by a potential of few tens of keVs. Since the energy is shared by the large gas-cluster ions, the dopant atoms obtain

only a small fraction of the total energy (i.e. of few eVs). The gas-cluster ion beam momentarily generates a high surface temperature and pressure, leading to an incorporation or infusion of the dopant atoms into the substrate. Since the cluster linearly and locally raises the substrate temperature, the stopping range for infusion doping follows the acceleration energy to the $1/3$ power. The process significantly suppresses the channeling effect and results in abrupt shallow junctions. High doping throughput can also be achieved due to the high number of dopant atoms infusion into the substrate. Unlike plasma doping that generates a plasma over the wafer, infusion doping has a cleaner process without metal contaminants from the side wall. The precise control of dopant dose by gas-cluster ion dose, however, remains challenging for this process.

2.3 Post-Implantation Annealing

2.3.1 Defect evolution upon annealing

Ion implantation generates a high concentration of interstitial dopant atoms and lattice defects. A post-implantation annealing is therefore necessary in providing a sufficient thermal energy for interstitial dopant atoms to move into substitutional sites and for substrate atoms to rearrange, repairing implantation-induced lattice defects. A number of defect structures are produced upon the initial annealing. Further annealing leads to the dissolution of these defect structures.

Ion implantation below the amorphization threshold inevitably generates extremely high amount of Frenkel pairs consisting of vacancies and interstitials in the Si substrate. Frenkel pairs undergo interstitial-vacancy recombination during the relaxation of the collision cascade. Due to a forward momentum of the interstitials, the excess interstitials

distribute deeply into the Si substrate while the excess vacancies reside in the surface region of the Si substrate. The “+1” model of Giles depicts a distribution of excess interstitials with a dose approximately equals to that of the implantation dose [54],[55]. It should be noted that small interstitial and vacancy clusters, point defect-dopant complexes, and amorphous pockets are possible upon the non-amorphizing implant.

The most unstable form of the excess interstitials is the single free interstitial that can readily diffuse or form a more stable structure (i.e. with vacancies, interstitials or dopant atoms) upon post-implantation annealing. Implantation dose greater than $1 \times 10^{12} \text{ cm}^{-2}$ but less than the amorphization threshold (after a low temperature annealing of 750°C), extended defects are found to be mainly of $\{311\}$ type. The $\{311\}$ defects consist of interstitial agglomerates located on $\{311\}$ habit plane and elongated in the $\langle 100 \rangle$ direction [56]–[58]. Since the $\{311\}$ defects have no dangling bonds along their side, the defects reduce the free energy of the excess interstitials. Strained constructed bonds, however, still exist at the both ends of the defects. It has been shown that transient enhanced diffusion (TED) is related to the dissolution of the $\{311\}$ defect [59],[60]. This leads to an enhanced boron diffusion of several orders greater than that predicted by the intrinsic diffusion [61].

For implantation dose greater than $1 \times 10^{14} \text{ cm}^{-2}$, both $\{311\}$ defects and dislocation loops may exist upon a post-implantation annealing. The dislocation loops are more stable than the $\{311\}$ defects. It is also found that the density of the dislocation loops increases with decreasing density of the $\{311\}$ defects upon further annealing. Both perfect prismatic loops and faulted loops are observed at moderate temperature annealing. On the other hand, only faulted dislocation loops are observed upon high temperature annealing [62]. Further

annealing leads to a coarsening of dislocation loops, i.e. growth of larger dislocation loops at the expense of smaller dislocation loops. This coarsening process can be described by the Ostwald ripening process, which involves a contribution of interstitials from the smaller dislocation loops to the larger dislocation loops. The size and density of both perfect and faulted dislocation loops decrease with prolonged annealing. It should be noted that the excess interstitials released from the dissolution of the dislocation loops could also produce enhanced dopant diffusion. The annealing temperature, however, is typically high (i.e. for the dissolution of the dislocation loops) that the extent of diffusion enhancement is not as large as that of the dissolution of the $\{311\}$ defects.

A different kind of defect evolution occurs in the case where the Si substrate was preamorphized. Preamorphization implantation (e.g. Si^+ or Ge^+ implantations) generates a highly damage region beyond the a/c interface. This EOR region contains a supersaturation of excess interstitials [37]–[39],[55]. Post-implantation annealing leads to the SPE regrowth of the amorphous layer. The annealing also causes the injection of excess interstitials into the bulk and into the surface region. It has been reported that the injection of interstitials into the boron-rich region causes the formation of the small BICs and leads to TED upon the release of excess interstitials [37]–[39]. Since the original amorphous layer recrystallizes into a perfect crystalline structure upon proper SPE process, the $\{311\}$ defects and the EOR loops are located in the EOR region. Due to a spatial separation between the EOR region and the dopant profile, the release of the excess interstitials from the dissolution of the EOR defects could also contribute to the enhanced dopant diffusion.

2.3.2 Dopant diffusion upon annealing

Fick's first law of diffusion described the macroscopic phenomena of dopant diffusion, which depends on the concentration gradient. Equation 2.2 depicts the Fick's first law of diffusion [28].

$$J = -D \frac{\partial C(x,t)}{\partial x} \quad (\text{Equation 2.2})$$

where J is the flux per unit area, D is the diffusion coefficient, $C(x,t)$ is the concentration profile, x is depth along the direction of the concentration gradient, and t is the time. Fick's second law, on the other hand, describe a time dependent diffusion characteristics and given below,

$$\frac{\partial C(x,t)}{\partial t} = D \frac{\partial^2 C(x,t)}{\partial x^2} \quad (\text{Equation 2.3})$$

Equation 2.3 assumes that the diffusion coefficient, D , is independent of time and space.

The intrinsic diffusion coefficient can be represented by an Arrhenius equation.

$$D = D_o e^{\frac{E_A}{kT}} \quad (\text{Equation 2.4})$$

where D_o is the pre-exponential factor, E_A is the activation energy of the diffusion process, k is the Boltzmann's constant, and T is the temperature.

As Fickian diffusion only describes the macroscopic diffusion characteristics, a complete picture regarding dopant diffusion requires insight in the atomistic diffusion characteristics. Dopant atoms may diffuse in Si lattice through (a) direct dopant exchange, (b) vacancy-dopant interaction, (c) interstitialcy-dopant interaction, and (d) interstitial-dopant interaction. Direct dopant exchange proceeds with a substitutional dopant atom exchanging position with an adjacent Si atom. This process requires a high activation energy associated with the breaking of Si bonds for the exchange to occur. Vacancy

assisted diffusion occurs when a substitutional dopant atom exchange position with a adjacent vacant lattice site. The activation energy for the diffusion of the vacancy-dopant pair is less than that of dopant atom diffusing by itself. As a result, the vacancy mechanism is more energetically favorable than that of direct dopant exchange. The interstitialcy and interstitial mechanisms are very similar, and it is difficult to distinguish the mechanisms by empirical methods. Dopant interstitialcy mechanism occurs when a dopant atom and a Si interstitial occupy a single lattice site. The dopant atom diffuses by translating positions with the nearest neighbors by bond exchange without displacing the lattice atoms. Dopant-interstitial mechanism occurs when a Si interstitial “kick-out” a substitutional dopant atom to form a dopant interstitial. The dopant interstitial diffuses through the interstices of the Si lattice until the dopant interstitial returns to a substitutional site through a “kick-in” mechanism or I-V recombination. Previous reports has shown that boron and phosphorus diffuse through the interstitial assisted mechanism, while arsenic exhibits both vacancy and interstitial diffusion [28],[61].

2.3.3 Transient enhanced diffusion

Post-implantation annealing can result in a significantly deeper boron profile due to TED [8],[9],[11]–[13],[63]–[67]. TED arises from the excess interstitials generated during ion implantation and causes boron diffusivity to be several orders larger than that of the equilibrium diffusivity. A Si interstitial (I) can react with a substitutional boron atom (B) to form a highly mobile B-I pair or to kick-out a boron atom to a mobile interstitial position [67]. The B-I pair or interstitial boron diffuses a considerable distance before the boron atom returns to a substitutional lattice.

The excess interstitials are injected during the dopant implantation or preamorphization implantations such as Si^+ and Ge^+ implantation. Despite recombination of Frenkel pairs upon relaxation of the collision cascade, excess interstitials of a dose equal to that of the implantation dose still exist due to a forward of momentum of the interstitials (i.e. the “+1” model) [54],[55]. The excess interstitials can form interstitial clusters and/or interstitial-boron clusters upon initial annealing. These metastable clusters or extended defects can contribute to TED upon their dissolution upon further annealing. Since the diffusion is driven by excess interstitials, the process is therefore transient which occurs until defects are dissolved or annihilated upon prolonged annealing.

It has been shown that TED is related to the amount of interstitials released from the $\{311\}$ defects [59],[60]. The $\{311\}$ defects grow by Ostwald ripening mechanism after nucleation. Further annealing leads to the dissolution of the $\{311\}$ defects, releasing excess interstitials for TED. The time necessary for the dissolution of the $\{311\}$ defects is on the order of minutes at 800 °C and on the order of seconds at 950 °C [2]. This coincides with the duration of TED. For a higher implantation dose, the $\{311\}$ defects may unfault into stable dislocation loops, which require a longer time to dissolve [59]. It is found that boron TED occurs in the absence of the $\{311\}$ defects, which suggest other source of excess interstitials for TED. It is believed that microscopic BICs are also the source of excess interstitials for TED in the case of high dose boron implantation [9],[37]–[39],[68]. The presence of electrically inactive and thermally immobile boron-interstitial clusters (BICs) leads to an immobile high concentration peak of the dopant profile, which does not diffuse upon initial annealing. These small BICs form in the region where boron and interstitial concentrations are high, typically at the peak of the boron profile. These clusters emit

excess interstitials for TED upon further annealing. It should be noted that the time required for the release of excess interstitials from BICs is much shorter than that from the {311} defects and dislocation loops upon the initial annealing.

The theoretical calculation for the activation energy of the increase in the junction depth indicates a negative value for the boron TED [69]–[71]. This implies that a high temperature annealing will induce a smaller extent of TED due to a high efficiency in defect annihilation as compared to that of a lower temperature annealing. The findings motivate the development of high temperature annealing with large ramp-up rate for the formation of ultra-shallow junctions.

2.3.4 Rapid thermal annealing

Conventional soak rapid thermal annealing (RTA) with a soak time of a few seconds and a moderate ramp-up rate faces tremendous challenges in the formation of ultra-shallow junctions. Boron TED induced by the soak RTA results in deep junctions unacceptable for nano-scale CMOS technology [24],[67]. It has been shown that high temperature RTA with large ramp-up rates, i.e. greater than 200 °C/s, reduces the extent of boron TED [8],[67],[72],[73]. Interstitial clusters and interstitial-boron clusters are not stable at high temperatures. Instead of inducing significant TED, the excess interstitials annihilate through other pathways, such as diffusion to the substrate surface or into the substrate bulk, recombination with vacancies, or pairing with other defects.

Spike RTA uses a less than 1 s soak time with a much higher ramp-up rate, as high as 400 °C/s [24]. Spike RTA can achieve a high temperature annealing (i.e. slightly greater than 1000 °C) with a relatively small thermal budget, which reduces the extent of TED.

However, Argawal *et al.* has shown that further increase in the ramp-up rate above a certain value produces insignificant reduction in TED. This is due to a limited RTA ramp-down rate, i.e. 70 – 90 °C/s [8]. Despite that spike RTA produces a higher activation through high temperature annealing, the dopant activation is still limited by the equilibrium solid-solubility of the dopant.

2.3.5 Flash annealing

Flash annealing typically consists of two annealing steps. An intermediate annealing raises the substrate temperature to a moderate temperature (e.g. 800 °C) at a ramp-up rate of 200 °C/s to 400 °C /s using a rapid thermal annealer. At the peak of the intermediate annealing, a flash lamp (or arc lamp) is used to raise the substrate surface to a high temperature (e.g. 1200 °C) in milliseconds regime [20],[74]–[76]. The millisecond flash annealing is achieved by discharging a capacitor bank into the flash lamp. Due to a reduced thermal mass, the flash lamp responds rapidly and minimizes the ramp-up and the ramp-down rates. The reduced thermal budget is capable of producing a highly activated junction with minimum diffusion. The flash lamp produces a spectral distribution that can readily be absorbed in Si. This is attributed to that over 95% of the radiation is below 1.2 µm (i.e. the band gap absorption of Si) [74]. Due to a high absorption at the substrate surface, the flash lamp is capable of increasing the substrate surface temperature greater than 1000 °C while increasing the substrate bulk temperature not more than 50 °C of the intermediate temperature. However, challenge still remains in the uniformity of this highly non-equilibrium annealing across the large substrate area. Moreover, the highly non-equilibrium annealing creates a high thermal stress to the substrate, leading to substrate deformity or

even breakage. Practical substrate consists of complex device structures that may also complicates the highly non-equilibrium annealing [77]. The geometry dependent annealing will adversely affect the activation and distribution of the annealed junction.

2.3.6 Extending rapid thermal annealing

Si^+ and Ge^+ implantation above the amorphization threshold generates a continuous amorphous layer at substrate surface [2]. This amorphous layer suppresses dopant channeling, especially for light dopant atoms such as boron, and reduces the as-implanted dopant profile depth. Post-implantation annealing causes a solid-phase epitaxial (SPE) regrowth of the amorphous layer using the crystalline template of the c-Si substrate at the a/c interface [20],[40],[67],[78]–[80]. SPE regrowth is typically conducted using a low temperature furnace or rapid thermal annealing at 500 °C to 600 °C. Dopant atoms are incorporated into the substitutional sites during the rearrangement of the Si atoms. This leads to a dopant activation greater than the dopant solid solubility with little dopant diffusion. However, Si^+ and Ge^+ preamorphization generates a high concentration of EOR damage beyond the a/c interface, which may lead to substantial diffusion to the dopant profile tail. If this residue damage is located near the junction depletion region, it can cause excessive leakage to the junction. Further annealing also causes a dopant deactivation to the already activated dopant due to a backfill of excess interstitials from the EOR region [37]–[39].

BF_2^+ implantation can also create a surface amorphous layer that reduces dopant channeling and enhances dopant activation through SPE regrowth. Furthermore, fluorine atoms suppress TED of boron atoms by trapping of Si interstitials [2],[37]. It has been

shown that co-implantation of fluorine leads to a shallow junction formation upon post-implantation RTA. Fluorine atoms also suppress dopant deactivation in the SPE annealing technique. However, fluorine atoms retard boron activation during post-implantation RTA due to the formation of B-F complexes. Fluorine atoms also enhance boron penetration in the gate region, causing a threshold voltage instability [42],[43].

Co-implantation of carbon has been shown to create a shallower dopant profile upon post-implantation RTA [2],[81],[82]. The reduced TED is attributed to the formation of Si-C complexes with the Si interstitials. It has been found that {311} and other extended defect can be reduced with carbon co-implantation. Carbon co-implantation, however, can cause high junction leakage due to the formation of Si-C precipitates.

Implantation below the amorphization threshold generates a high concentration of Frenkel pairs of vacancies and interstitials. Due to a forward momentum of the interstitials, the excess interstitials distribute deeply into the Si substrate [83]–[86]. This creates a high concentration of excess vacancies near the surface region of the substrate. This vacancy-rich region readily incorporates dopant atoms, leading to a high dopant activation upon post-implantation RTA. This technique requires a MeV implantation of Si^+ or Ge^+ prior to boron implantation, which leads to the spatial separation of excess vacancies and interstitials. By placing boron atoms in the vacancy-rich region further away from the interstitial-region, boron TED can be suppressed. The MeV implantation inevitably creates a highly damaged region (i.e. the high concentration of excess interstitials) in the bulk of the substrate. This damaged region should be located far from the junction region, resulting in minimum effect on the junction leakage [83].

2.4 Laser Annealing

2.4.1 Excimer lasers

Excimer lasers operate on the radiative transition between a molecular excited state and a molecular ground state of gas molecule. Excimer lasers typically use gas molecules formed from noble gases (e.g. Ar, Xe and Kr) and halogens (e.g. F, Cl and Br). More than 90% of the gas mixture is a buffer gas (i.e. another rare gas) that does not take part in the chemical reaction but to assure momentum conservation. The rare gas halide lasers can be excited by electron beam excitation or pulsed electrical discharge to form the excited molecules, $(A^+X)^*$, which is the upper state population. The excited molecules has an extremely short lifetime and drop to ground state, AX , by releasing photons. Therefore, excimer lasers operate in tens of nanoseconds pulse and tend to have shorter wavelength with higher efficiency than most conventional lasers. Pulsed electrical discharges are more commonly used in commercial lasers for higher pumping efficiency and higher average power [87]. The primary reactions excited by electrical discharge is given by

where A is the rare gas ion, X is the halogen ion, and M is another rare gas ion to assure momentum conservation.

2.4.2 Laser irradiation interaction with silicon

Si is an indirect band gap material, of which the valence band maximum and conduction band minimum locates at different momentum, i.e. wave vector k . For absorption of a photon with $h\nu \approx E_g$, the simultaneous absorption or emission of a phonon

providing extra momentum is required [88],[89]. Such a process is known as indirect transition and may occur in two ways [90]. The first way involves an electron in the valence band maximum absorbs a photon and makes a transition to an intermediate state in the conduction band of a same wave vector. A lattice phonon is then absorbed and the electron transits to the conduction band minimum. The second way occurs when a photon excite an electron from a valence band maximum directly below the conduction band minimum with a hole being transferred to the valence band maximum by phonon absorption. Direct transition is also possible in Si, provided that the photon energy is sufficiently large, e.g. $h\nu > 3.4$ eV for Si. The relaxation of the excited electron back to the valence band emits a phonon [90]. This phonon emission process results in a transfer of energy to the lattice, which raises the lattice temperature. This transfer of energy from the photon to the lattice occurs in a time much shorter than the laser pulse duration. It can be assumed that the laser irradiation is absorbed and converted into thermal energy instantaneously [88],[89].

Since optical wavelengths, λ , are typically larger than the atomic distance, the response of a homogeneous material can be described in terms of averaged macroscopic quantities. It is convenient to describe the response of material to the light wave in terms of reflectivity, R , and the absorption coefficient, α . The two quantities is given by [88]

$$R = \frac{(n-1)^2 + \kappa^2}{(n+1)^2 + \kappa^2} \quad (\text{Equation 2.6})$$

$$\alpha = \frac{4\pi\kappa}{\lambda} \quad (\text{Equation 2.7})$$

where n is the real part of the complex reflective index and κ is the imaginary part of the complex reflective index.

Figure 2.1 gives the absorption coefficient and reflectivity of c-Si as a function of wavelength at room temperature. It is found that α is in the order of 10^6 cm^{-1} when $\lambda < 360 \text{ nm}$. This is several orders of magnitude greater than that for $\lambda > 540 \text{ nm}$. It is therefore desirable to use lasers with $\lambda < 360 \text{ nm}$ for efficient heating of the Si substrate by the optical absorption.

Figure 2.1. Absorption coefficient and reflectivity of c-Si as a function of wavelength at room temperature. Redrawn from Ref. 88.

2.4.3 Heat flow calculations

Heating in an laser irradiated Si substrate is a consequence of the balance between the deposited energy, which is governed by the optical parameters of the substrate, and the heat diffusion, which is determined by the thermal parameters of the substrate and laser pulse duration [88],[91]. Since the transfer of energy from the photon to the lattice occurs in a time much shorter than the laser pulse duration, it can be assumed that the

laser light is converted into heat instantaneously. Taking that the laser beam travels along z-axis and the composition is homogeneous in the x, y plane, the one-dimensional heat equation can be expressed as

$$\frac{\partial T(z, t)}{\partial t} = \frac{\alpha}{\rho(T)C(T)}(1-R)I_o(t)e^{-\alpha z} + \frac{1}{\rho(T)C(T)}\frac{\partial}{\partial z}\left[K(T)\frac{\partial T(z, t)}{\partial z}\right] \quad (\text{Equation 2.8})$$

where z is depth, t is time, T is temperature, $\rho(T)$ is density, $C(T)$ is specific heat capacity, $K(T)$ is thermal conductivity, R is reflectivity of sample, α is absorption coefficient of sample, and $I_o(t)$ is laser power intensity upon reaching sample surface.

The laser induced heating process is affected by the absorption length, α^{-1} , and the heat diffusion length, $L_T = \sqrt{2D_h t_p}$ where $D_h = \frac{K(T)}{\rho(T)C(T)}$ is the thermal diffusivity and t_p is the laser pulse duration. In the cases where α^{-1} is smaller than L_T , the laser absorption takes place on the surface region. The average increase in the surface temperature can be simplified as

$$\Delta T(t) \approx \frac{(1-R)I_o(t)}{K} \sqrt{\frac{D_h t_p}{2}} \quad (\text{Equation 2.9})$$

In the case where L_T is smaller than α^{-1} , the average increase in the surface temperature is given by

$$\Delta T(t) \approx \frac{(1-R)I_o(t)t_p \alpha e^{-\alpha z}}{\rho C} \quad (\text{Equation 2.10})$$

For a 248 nm laser irradiation with $t_p = 23$ ns on a c-Si substrate, α^{-1} is about 5.6 nm while L_T is about 2 μm (i.e. $\alpha^{-1} < L_T$). The conduction of heat in the c-Si substrate plays an important role during laser irradiation [88].

2.4.4 Dopant incorporation upon LA

LA typically melts the surface region of the implanted substrate and causes the dopant to be distributed uniformly, producing a box-like profile after a liquid-phase epitaxial recrystallization of the melted region. This is possible due to the fact that the diffusivity of boron in liquid Si is about eight orders higher than that in solid Si [22],[24]. This leads to the formation of highly activated ultra-shallow abrupt junctions determined by laser induced melt depth. The highly non-equilibrium LA produces a high concentration of electrically activated dopant that exceeds the solid solubility limit [88],[89]. The enhanced dopant activation can be explained by solute trapping upon recrystallization of the melted region [88],[89]. When the time required to regrow one or more monolayer of atoms during solidification process is significantly shorter than the residence time of the dopant at the interface, the dopant has a high probability of being incorporated into the growing solid. This leads to exceedingly high amount of dopant atoms being incorporated into the substitutional sites during the solidification.

It is observed that the box-like dopant profile is attributed to a near unity interface segregation coefficient, k_i , of which k_i is defined as the ratio of the solute concentration in the solid to that in the liquid at the liquid/solid interface. The equilibrium k_i of common dopants are lower than unity (e.g. 0.8 for boron and 0.35 for phosphorus) [89], which would produce graded dopant profile upon solidification. Wood *et al.* have indicated that the k_i increases from an equilibrium value with elevated recrystallization rate and reaches unity at high recrystallization rate, as shown in figure 2.2 [89]. This gives a good fit between the simulated dopant profile and empirical dopant profile.

LA can also be conducted in a non-melt regime, in which the laser fluence used is below the threshold of melting of the substrate. The laser irradiation only heats up the c-Si substrate to a temperature below the melting temperature of c-Si substrate [26],[27]. The near zero thermal budget of non-melt LA induces a negligible diffusion to the as-implanted dopant profile. The extremely rapid heating and quenching also leads to a high dopant incorporation due to dopant trapping.

Figure 2.2. Segregation coefficients of common dopants as a function of melt front velocity. Redrawn from Ref. 89.

2.5 Summary

The background and relevant theories on the formation of ultra-shallow junctions has been discussed. The relevant information and results on the interaction between laser irradiation and materials have also been covered. Further discussion with regard to the theories will be made in the following sections.

CHAPTER 3

EXPERIMENTAL DETAILS

3.1 Introduction

This chapter describes the experimental setup, sample preparation, materials and electrical characterization techniques used in this project. Definitions and determinations of junction depth and melt depth from secondary ion mass spectrometry are also given in this chapter.

3.2 Ion Implantation

B^+ was implanted into 200 mm Czochralski n-type (100) silicon substrates, silicon-on-insulator (SOI) substrates and strained-Si/SiGe substrates. Figure 3.1 shows the ultra-low energy B^+ implantation conditions used in this project. The ultra-low energy B^+ was implanted at energies of 0.2 keV, 0.5 keV or 1.0 keV to doses of $2.5 \times 10^{14} \text{ cm}^{-2}$, $5 \times 10^{14} \text{ cm}^{-2}$ or $1 \times 10^{15} \text{ cm}^{-2}$. Figure 3.2 shows the as-implanted boron profiles at energies of 0.2 keV, 0.5 keV or 1.0 keV to a dose of $5 \times 10^{14} \text{ cm}^{-2}$ in a c-Si sample. The boron profiles were obtained using secondary ion mass spectrometry (SIMS), of which the detailed setup will be discussed in section 3.5.1. As shown in figure 3.2, lowering the implantation energies creates a shallower as-implanted boron profile. Since the B^+ was implanted into the c-Si sample, significant boron profile tail is generated through ion channeling even at ultra-low energy as low as 0.2 keV.

Figure 3.1. Ultra-low energy B⁺ implantation conditions used in this project.

Figure 3.2. As-implanted boron profiles at energies of 0.2 keV, 0.5 keV or 1.0 keV to doses of $5 \times 10^{14} \text{ cm}^{-2}$ in a c-Si sample. The boron profiles were obtained using secondary ion mass spectrometry (SIMS), of which the detailed setup will be discussed in section 3.5.1.

Figure 3.3. (a) PAI depth induced by Si⁺ or Ge⁺ implantations. Cross-sectional transmission electron microscopy (TEM) micrographs of original c-Si samples upon Si⁺ PAI with a dose of $2 \times 10^{15} \text{ cm}^{-2}$ at (b) 10 keV, (c) 20 keV and (d) 30 keV.

Prior to the B⁺ implantation, some of samples were preamorphized using Si⁺ or Ge⁺ implantations. Figure 3.3 (a) shows that the Si⁺ or the Ge⁺ preamorphization implantations (PAI) create a continuous amorphous layer from the substrate surface which increases with the increasing implantation energy. The PAI depths shown in figure 3.3 (a) were obtained

using cross-sectional transmission electron microscopy (TEM). Details on the TEM setup will be discussed in section 3.5.3. Figures 3.3 (b) to (d) show the corresponding cross-sectional TEM micrographs of the original c-Si samples upon a Si⁺ PAI with a dose of $2 \times 10^{15} \text{ cm}^{-2}$ at 10 keV, 20 keV and 30 keV, respectively. The Si⁺ implantation at 10 keV and with a dose of $1 \times 10^{14} \text{ cm}^{-2}$, on the other hand, lies below the implantation threshold required to generate a continuous α -Si layer [2]. Despite that the Si⁺ subamorphization implantation (SAI) does not generate the continuous α -Si layer, the implantation creates excess interstitials, vacancies and amorphous pockets.

Figure 3.4. As-implanted boron profiles at energy of 0.5 keV to a dose of $1 \times 10^{15} \text{ cm}^{-2}$ in a c-Si and a Si⁺ PAI (i.e. 10 keV and $2 \times 10^{15} \text{ cm}^{-2}$) samples.

Figure 3.4 shows the as-implanted boron profiles at 0.5 keV with a dose of $1 \times 10^{15} \text{ cm}^{-2}$ in the c-Si and the Si⁺ PAI samples. The Si⁺ PAI was conducted at 10 keV and $2 \times 10^{15} \text{ cm}^{-2}$, creating a continuous α -Si layer of 30 nm (also refer to figure 3.3 (b)). It is

observed that the 30-nm α -Si layer effectively suppresses the ion channeling, leading to a shallower dopant profile tail.

3.3 Simulation Studies

The heating and melting characteristics induced by a KrF 248-nm excimer laser were simulated using the simulation of laser interaction with materials (SLIM) [92]. SLIM uses an implicit finite-difference method to simulate the interaction of an intense nanosecond laser beam with the substrate materials. It provides a numerical solution to the one-dimensional heat flow equation as described in section 2.4.3. Simulation conducted in this project assumes an initial substrate temperature at room temperature and the pulse shape to be trapezoidal with a full-width at half-maximum (FWHM) of 23 ns. The thermal and optical parameters of the c-Si and the α -Si substrates used for the simulation are listed in tables 3.1.

Simulation of implantation cascades were performed using the transport of ions in matter (TRIM) [93] and UT-MARLOWE [94]. TRIM calculates the stopping and range of ions into a matter based on a quantum mechanical treatment of ion-atom collisions. As TRIM assumes ions implantation into an amorphous substrate, the simulation predicts a shallower dopant profile than that in a crystalline substrate. UT-MARLOWE, similarly, is based on Monte Carlo computer simulations using a binary collision approximation (BCA). It has been shown that UT-MARLOWE produces certain accuracy in the simulation of ion implantation into a crystalline silicon substrate [95]. In addition, the cumulative damage models in UT-MARLOWE produces reasonable estimates of

interstitials and vacancies distribution upon ion implantation. Other simulation tools used in this project will be discussed in subsequent texts.

Table 3.1. Thermal and optical properties of c-Si and α -Si [96].

Properties	c-Si	α -Si
Melting temperature (K) [97],[98]	1688	1488
Latent heat of melting (J/cm ³) [98]–[100]*	4206	3076
Thermal conductivity (W/cmK) [98]–[100]*	$1585T^{-1.229}$ (T < 1370 K) 0.221 (1370 K < T < 1688 K) 1.4 (T > 1688 K)	0.018 (T < 1688 K) 1.4 (T > 1688 K)
Volume heat capacity (J/cm ³ K) [97],[99],[101]*	$2.54 \times 10^{-4}T - 3.68 \times 10^{-4}T^{-2} +$ 1.99 (T < 1688 K) 2.49 (T > 1688 K)	$2.64 \times 10^{-4}T - 3.82 \times 10^{-4}T^{-2} +$ 2.06 (T < 1688 K) 2.49 (T > 1688 K)
Reflectivity [88],[102]*	0.67 (T < 1688 K) 0.70 (T > 1688 K)	0.55 (T < 1688 K) 0.70 (T > 1688 K)
Absorption coefficient (cm ⁻¹) [102],[103]*	1.81×10^6 (T < 1688 K) 1.93×10^6 (T > 1688 K)	1.40×10^6 (T < 1688 K) 1.93×10^6 (T > 1688 K)

*derived or calculated from the references.

3.4 Laser Annealing

The laser annealing was performed with a 248 nm KrF excimer laser (Lamda Physik LPX220i or Novaline). The 248 nm excimer laser has a pulse duration of 23 ns at a repetition rate of 1 Hz. The laser beam was directed by a mirror and lens system before reaching the sample surface, as shown in figure 3.5. To ensure good uniformity of the laser beam reaching the sample surface, the laser was equipped with a beam homogenizer forming a $3 \times 3 \text{ mm}^2$ spot on the sample. The laser beam was scanned on the sample surface across the x and y directions, enlarging the annealed area to greater $1 \times 1 \text{ cm}^2$. The process was conducted in normal or nitrogen ambience at room temperature.

Figure 3.5. A schematic diagram of the equipment setup for LA.

3.5 Materials Characterization

3.5.1 Secondary ion mass spectrometry

SIMS involves a primary ion beam impinging on the sample surface, which causes atoms or molecules from the sample to be sputtered off. The ionized atoms or molecules from the sample are then detected and analyzed by a mass spectrometer. The sequential sputtering of atoms or molecules from the sample enables the depth profiling of certain

elemental or molecular species. The SIMS dopant profiles in this project were analyzed using a Cameca IMS 6f instrument with high-resolution magnetic sector mass spectrometer. The instrument was equipped with a secondary ion post-acceleration system for enhancing electron multiplier (EM) signals at low primary ion energies. Positive SIMS analyses (i.e. the detection of positive secondary ions) were conducted using a primary beam of O_2^+ ions with a beam current of 10 to 30 nA and at a net energy of 1 keV. The primary ions impinged the surface at an incident angle of 56° , with respect to the surface normal. Negative SIMS analyses (i.e. the detection of negative secondary ions) used a primary beam of Cs^+ ions with a beam current of 20 nA and net energy of 15 keV. Typical depth profiling was performed over a scanned area of $250 \times 250 \mu m^2$. The crater depths were measured using a Tencor Alpha-Step 500 stylus profilometer.

3.5.2 Determination of junction depth, melt depth and junction abruptness from SIMS profiles

As the diffusivity of boron in the liquid phase Si is about eight orders higher than that in the solid phase Si, dopant atoms can diffuse uniformly within the melt region forming an abrupt dopant profile. However, dopant profile obtained by SIMS normally exhibits a sloping trailing tail that is limited by the SIMS depth resolution. Ion beam mixing and sputtering induced roughness limit the SIMS depth resolution, which leads to inaccurate representation of the actual dopant profile [104]. Figure 3.6 depicts a typical SIMS dopant profile after LA, in which two distinct portions can be distinguished, namely an initial flat region and a steep trailing tail. As that the junction is abrupt, the location of one decade lower than where the slope experiences a sharp change can be

defined as the melt depth (x_m). This definition takes into the consideration of the ion mixing effect and sputtering induced roughness that limit the SIMS depth resolution. However, in the case of graded junction, the melt depth (x_m) cannot be easily determined. Junction depth (x_j) defined at the background concentration of $1 \times 10^{18} \text{ cm}^{-3}$ remains a good determinant for the dopant profile. Notations of x_m as the melt depth and x_j as the junction depth are both used in the rest of the chapters for discussion of the LA induced junction features. Junction abruptness is defined as the slope of the abrupt trailing edge of the SIMS profile. As dopant profile obtained by SIMS normally exhibits a sloping trailing tail that is limited by the SIMS depth resolution, a better actual junction abruptness is expected when the extracted junction abruptness is equal to the SIMS depth resolution.

Figure 3.6. Definition and determination of the junction depth, melt depth and junction abruptness from SIMS profile.

3.5.3 Transmission electron microscopy

TEM analysis is obtained by bringing accelerated electrons into focus on a sufficiently thin sample. The sample has to be thinned to electron transparency that does not cause spreading of the incident electron beam. The transmitted and forward scattered electrons form a diffraction pattern in the back focal plane and a magnified image in the image plane. Microstructural and lattice information of the as-implanted and annealed samples in this project were obtained using cross-sectional TEM analyses. The samples were first thinned to electron transparency using the standard mechanical polishing and low-energy ion milling. The samples were then examined under a Philips Tecnai F20 transmission electron microscope, which operates at an accelerating voltage of 200kV.

3.5.4 Atomic force microscopy

Atomic force microscopy (AFM) was employed to characterize the surface morphology of the as-implanted and the annealed samples. No chemical treatment for removal of native oxide was performed to prevent loss of surface feature after cleaning. The measurements were conducted in normal ambience with a tapping mode. The tapping mode etched silicon probe (TESP) tips were used for the analyses. The cantilever was oscillated near its resonant frequency as it scanned over the sample surface to obtain high-resolution images without inducing destructive friction forces. The images were obtained by scanning over an area of $1 \times 1 \mu\text{m}^2$ with a resolution of 512×512 data points.

3.5.5 X-ray diffraction

X-ray diffraction (XRD) analysis was employed to determine the phase formation in the Ni-silicided samples. The XRD diffractograms consist a series of different maxima or peaks at specific Bragg angles, θ . Each peak corresponds to crystallographic planes in the crystal that causes the respective constructive interference of the diffracted x-ray. The interplanar spacing (d_{hkl}) and lattice parameter (a) can be obtained from the XRD peak position by the following relations [105],

$$n\lambda = 2d_{hkl} \sin \theta \quad (\text{Equation 3.1})$$

$$d_{hkl} = \frac{a}{\sqrt{h^2 + k^2 + l^2}} \quad (\text{Equation 3.2})$$

where λ is the wavelength of the incident x-ray, n is an integer and h , k and l are the Miller indices of the crystal. This provides the information on the crystal structure of the materials. A Cu x-ray source at an accelerating voltage of 45 kV and at a current of 40 mA at a grazing incidence of 0.5° was used in this project. The diffracted x-ray beam from the sample was scanned from 20° to 80° (i.e. 2θ).

3.6 Electrical Characterization

3.6.1 Four-point probe measurement

Sheet resistance is typically used to describe the resistivity of thin films in semiconductor industry. It is proportional to the resistivity ρ of the thin film and is inversely proportional to its thickness t . The sheet resistance is then given as,

$$R_s = \frac{\rho}{t} \quad (\text{Equation 3.3})$$

The four-point probe technique is a quick and common method for measuring sheet resistance. As shown in figure 3.7, the current, I , is typically carried through the outer two contacts into the sample while the potential difference, V , is measured across the inner contacts. For a sample with arbitrary dimensions, the sheet resistance is given by

$$R_s = F \frac{V}{I} \quad (\text{Equation 3.4})$$

where F is a correction factor that depends on the sample geometry, sample thickness, probe spacing, and probe location. For a sample that is sufficiently bigger than the probe spacing and can be assumed to be infinite in all planar directions, F has a value of 4.532. The four-point probe measurement in this project used a probe spacing s of 1 mm.

Figure 3.7. Four-point probe with probe spacing s for sheet resistance measurement.

The downward pressure of the 4 probes on the sample surface may penetrate the highly doped ultra-shallow junction during the four-point-probe sheet resistance measurement. This may create an additional current pathway in the bulk of the substrate during the sheet resistance measurement, adding the resistivity information of the bulk of the substrate to the highly doped ultra-shallow junction resistivity. It should be noted that

the resistivity of the substrate used in the study is about 4 orders higher than that of the highly doped ultra-shallow junction. This leads to a negligible measurement current into the bulk of the substrate. The measurement current is therefore mainly confined in the highly doped ultra-shallow junction, which results in sheet resistance information reflecting the ultra-shallow junction. Furthermore, the sheet resistance obtained using Hall measurement (which will be further discussed in the following section) has also been compared with the four-point-probe measurement and the difference is within the 10% of the experimental tolerance.

3.6.2 Hall measurement

Hall effect measurement provides electrically active carrier concentration, carrier mobility and sheet resistance of thin films. Figure 3.8 shows a typical hall effect measurement setup, in which the current, I , is carried through two diagonal contacts into the sample while an orthogonal magnetic field, B , is applied. Free carriers experience a Lorentz force and drift away perpendicularly from the current path, leading to an excess surface charges. This results in a transverse Hall voltage, V_H , and is given by,

$$V_H = \frac{IBR_H}{t} \quad (\text{Equation 3.5})$$

where R_H is the Hall coefficient and t is the sample thickness. The sheet carrier concentration n_s can then be calculated from the measurement as follows,

$$n_s = \frac{IB}{q|V_H|} \quad (\text{Equation 3.6})$$

where q is the elementary charge. Hall effect measurement in this project used a square sample geometry of $1 \times 1 \text{ cm}^2$. Ohmic contacts were formed by In metal on the corners of

the sample cleaned by standard diluted HF solution. The magnetic field was fixed at 0.32 T and the measurement were conducted at room temperature.

Figure 3.8. Hall effect measurement with In contacts.

3.6.2 I-V characterization

The current-voltage (I-V) characteristics were measured from p^+/n diodes and pMOSFETs using HP4156B semiconductor analyzer. The I-V characterization on the junction leakage current of the p^+/n diodes provides information on the residual defect annihilation after LA. This is due to challenges in the residual defect annihilation by nano-second laser irradiation in obtaining a defect free ultra-shallow junction using LA. The p^+/n diodes and pMOSFETs were fabricated by a conventional CMOS process flow. Unless otherwise stated, the implantations used on the fabricated devices follows that in the unpatterned sample. It should be noticed that an optimal laser fluence obtained on a unpatterned sample or on a patterned devices may vary due to the presence of oxide isolation and gate stacks [106]–[109]. A sufficiently wide range of laser fluences was used for the study to confirm the consistent results observed in the unpatterned and the device

samples. To avoid additional thermal process, no silicidation was performed on most of the devices.

3.7 Summary

Various topics on the experimental setup, sample preparation, materials and electrical characterization techniques used in this project are briefly discussed. Unless otherwise stated, the subsequent chapters follow the experimental procedure described in this chapter.

CHAPTER 4

FORMATION OF ULTRA-SHALLOW JUNCTIONS USING LASER ANNEALING

4.1 Introduction

This chapter studies and discusses the formation of ultra-shallow p^+/n junctions in silicon substrate upon LA. It begins with the comparison of the formation of ultra-shallow p^+/n junctions in crystalline and preamorphized silicon substrates upon single-pulse LA. The discussion includes the analyses of the junction depth, the junction abruptness, the sheet resistance, the dopant activation and the dopant loss in both structures when subjected to single-pulse LA. The later part of this chapter shows the effects of multiple-pulse shallow-melt and non-melt LA on the dopant activation, dopant loss and surface morphological modification of boron implanted crystalline silicon substrate. The last part of this chapter shows the enhanced dopant activation using a method based on Si^+ subamorphization implantation and non-melt LA.

4.2 Dopant Retention and Electrical Activation Induced by Single-Pulse LA in Boron Implanted Crystalline and Preamorphized Si Substrates

B^+ was implanted at 0.5 keV with a dose of $1 \times 10^{15} \text{ cm}^{-2}$ on an n-type Si (100) substrate. Prior to the B^+ implantation, some of wafers were preamorphized using Si^+ implantation at 10 keV with a dose of $2 \times 10^{15} \text{ cm}^{-2}$. Cross-sectional transmission electron

microscopy (XTEM) shows that a continuous amorphous layer of about 30 nm was created by the Si^+ implantation, as shown in figure 4.1.

Figure 4.1. Cross-sectional TEM micrograph of a B^+ implanted Si substrate that was preamorphized with a Si^+ implantation at 10 keV and $2 \times 10^{15} \text{ cm}^{-2}$.

Figures 4.2 (a) and (b) show the SIMS profiles of boron in the crystalline silicon (c-Si) and the preamorphized silicon (PAI-Si) samples, respectively, subjected to single-pulse LA at laser fluences of 0.5 and 0.6 J/cm^2 . The most prominent feature of figure 4.2 (a) is the formation of a box-like dopant profile in the c-Si sample upon (shallow-melt) LA at 0.6 J/cm^2 . The formation of the abrupt junction follows the melting of the surface region of the c-Si sample upon laser irradiation. Since the diffusivity of boron in liquid-phase silicon ($\sim 10^{-4} \text{ cm}^2/\text{s}$) is about eight orders of the magnitude higher than that in solid-phase silicon ($\sim 10^{-12} \text{ cm}^2/\text{s}$ at 1000 °C) [22],[24],[88], boron atoms redistribute uniformly within the melted silicon region. Due to the large difference in the diffusivity of boron in liquid- and solid-phase silicon, boron atoms have negligible diffusion beyond

the liquid/solid interface upon the extremely short duration of melting of silicon (i.e. in tens of nanoseconds) [22],[24]. This creates an abrupt junction that is defined by the laser induced melt depth, as shown in figure 4.2 (a). On the other hand, (non-melt) LA at 0.5 J/cm^2 does not induce any substantial melting in the c-Si sample [26],[27] and leads to a diffusionless dopant profile that follows the as-implanted profile, as shown in figure 4.2 (a). Since the laser irradiation at 0.5 J/cm^2 only heats up the substrate to a temperature below the melting temperature of silicon, the near-zero thermal budget of LA generates a negligible solid-phase boron diffusion.

Figure 4.2. SIMS profiles of boron in (a) crystalline and (b) preamorphized Si substrates subjected to single-pulse LA at laser fluences of 0.5 and 0.6 J/cm^2 . The 0.5 J/cm^2 laser annealed boron profile in (a) follows the as-implanted profile due to negligible diffusion upon laser irradiation.

In contrast to the case of the c-Si sample, similar box-like profiles are formed with LA at both 0.5 and 0.6 J/cm^2 in the PAI-Si sample, as depicted in figure 4.2 (b). This is

possible due to the fact that the melting temperature of amorphous silicon (α -Si) is about 200 °C lower than that of c-Si [22],[24],[110]. As a result, LA melts the α -Si layer without melting the underlying c-Si substrate. Since the melting in the PAI-Si sample is defined by the Si^+ implantation generated α -Si layer, abrupt ultra-shallow junctions of similar depth can be achieved in a wider range laser fluences. This leads to an improved process window of about 20% laser fluence.

Figure 4.3. High-resolution cross-sectional TEM micrographs of (a) crystalline and (b) preamorphized Si substrates subjected to single-pulse LA at 0.6 J/cm².

Figures 4.3 (a) and (b) show the high-resolution cross-sectional TEM micrographs of the c-Si and the PAI-Si samples, respectively, subjected to the single-pulse LA at 0.6 J/cm² shown in figure 4.2. The TEM micrographs indicate a perfect recrystallization process throughout the melted regions in both c-Si and PAI-Si samples. There is virtually no secondary defect observed in the liquid-phase epitaxial regrown regions [25]. Liquid-phase epitaxy begins at the liquid/solid interface at the maximum melt depth and

resolidifies using the crystalline template provided by the solid silicon. Despite the melted silicon underwent a rapid solidification rate of about 4 m/s ~ 8 m/s (simulated) in the PAI-Si and the c-Si samples, the resolidification rate is far below the resolidification rate required for α -Si formation (i.e. 15 m/s ~ 20 m/s) [110]–[112]. The melted silicon, therefore, has sufficient time for c-Si resolidification. It should be noted that the entire original α -Si layer (see figure 4.1) in the PAI-Si sample was melted in order to generate a virtually defect-free crystalline structure.

Figure 4.4. Junction depth, determined at a boron concentration of $1 \times 10^{18} \text{ cm}^{-3}$, and junction abruptness taken from the SIMS profiles in figure 4.2. The junction depth and the junction abruptness annealed at fluences greater than 0.6 J/cm^2 are also included.

Figure 4.4 shows the junction depth, determined at a boron concentration of $1 \times 10^{18} \text{ cm}^{-3}$, and junction abruptness, defined as the slope of the abrupt SIMS trailing profile, extracted from the SIMS profiles in figure 4.2. The junction depth and the junction

abruptness annealed at fluences greater than 0.6 J/cm^2 are also included. The junction depths in the c-Si sample remain at 32 nm when the laser fluence increases from 0.5 J/cm^2 to 0.6 J/cm^2 (i.e. from non-melt LA to shallow-melt LA). This is due to a melting shallower than 32 nm at 0.6 J/cm^2 , leading to a junction depth defined by the unannealed dopant profile tail. Upon a LA at fluences greater than 0.6 J/cm^2 , the junction depth in the c-Si sample shows a linear increase with increasing laser fluences. The linear increase in the junction depth with increasing laser fluences is attributed to a linear relationship of maximum melt depth in the c-Si sample with laser fluence, as shown in figure 4.5.

The maximum melt depth shown in figure 4.5 was extracted using the method described in section 3.5.2, while the simulated maximum melt depth was obtained using a one-dimensional heat flow calculation (i.e. SLIM) [92],[96]. Both extracted and simulated maximum melt depths indicate a threshold of melting of 0.52 J/cm^2 that defines the non-melt regime (i.e. region A) and shallow-melt regime (i.e. region B). The maximum melt depth follows a linear relationship upon laser irradiation at fluence greater than the threshold of melting. Upon a deeper melting induced at higher fluences, the dopant atoms were redistributed to a deeper extent, leading to a junction depth seen in figure 4.4. The junction abruptness in the c-Si sample remains approximately at 6 nm/decade up to a laser fluence of 0.73 J/cm^2 . SLIM indicates that laser irradiation at 0.75 J/cm^2 induces a deep melting of 85 nm (see figure 4.5) and a melt duration of 12 ns at the full-width-half-maximum (see figure 4.6 (a)). The short melt duration is insufficient to redistribute the boron atoms to the maximum melt depth achieved at 0.75 J/cm^2 . This leads to the formation of a less abrupt junction, as depicted in figure 4.4.

Figure 4.5. Extracted maximum melt depth (from SIMS profiles in figure 4.2 and profiles described in figure 4.4) and (SLIM) simulated maximum melt depth as a function of laser fluence.

The PAI-Si sample shows a constant junction depth at 0.5 J/cm^2 and 0.6 J/cm^2 , in which the melting is controlled by the original PAI depth as shown in figure 4.4. Upon a deeper melting at higher laser fluences, the junction depth in the PAI-Si sample increases with increasing laser fluences. As shown in figure 4.5, melting of the entire original α -Si layer occurs in the PAI controlled region (i.e. region I), leading to box-like junction profiles defined by the original α -Si layer. Overmelting beyond the original α -Si layer, which occurs at a higher fluence, sees a c-Si substrate, namely in the PAI limited region (i.e. region II) and the deep overmelting (i.e. region III) shown in figure 4.5. Overmelting in the PAI limited region melts slightly beyond the original α -Si layer, in which the lower melting temperature of the α -Si layer and high reflectivity of the melted Si surface limit

further melting. Deep melting beyond the original α -Si layer surpasses the effects of the α -Si layer and causes a melting similar to the c-Si sample, as depicted in figure 4.5 and figure 4.6 (b) at 0.75 J/cm^2 .

Figure 4.6. Simulated melt front upon laser irradiation of 23 ns FWHM starting at $t = 0$ on (a) crystalline and (b) preamorphized Si substrates as a function of time at different laser fluences.

The details of the SLIM simulated melt front profiles for the c-Si and the PAI-Si samples upon LA are shown in figures 4.6 (a) and (b), respectively. Upon laser irradiation (of 23 ns FWHM starting at $t = 0$), the surface temperature of the Si substrate builds up with respect to time. The surface temperature continues to rise and reaches the melting temperature of the Si substrate. (The increase on the surface temperature up to the melting temperature is represented in the duration of zero melt depth from time zero to time prior to any melt depth.) Upon reaching the melting temperature of c-Si sample, as depicted in figure 4.6 (a), a melt front propagates into the Si substrate. This is observed

as an increase in the melt depth with respect to time. Upon reaching the maximum melt depth, the melt front changes direction and solidifies towards the c-Si sample surface, which is observed as a decrease in the melt depth with respect to time. The derivatives of the melt depth to time can be used to describe the melting and resolidification rates. The melting (increasing edge of the curves) shows a high melting velocity approximately 4 m/s to 8 m/s. The resolidification (decreasing edge of the curves) shows a high resolidification velocity approximately 6 m/s to 8 m/s. The PAI-Si sample, on the other hand, shows a sharp increase in the melting of the 30-nm α -Si layer and remains at a constant melt front at the original amorphous/crystalline (a/c) interface before resolidification takes place at LA of 0.6 J/cm^2 . The flat melting profile is due to the lower melting temperature of the α -Si as compared to that of the c-Si [96],[110]. In this case, the temperature of the melted region maintains at a temperature slightly above the melting temperature of the α -Si but below the c-Si. When subjected to a higher fluence LA at 0.75 J/cm^2 , the temperature of the PAI-Si sample raises above the melting temperature of the α -Si and melts the original α -Si layer. Upon reaching the a/c interface, the temperature holds at the melting temperature of the α -Si until the absorbed energy raises the temperature further above the melting temperature of the c-Si, leading to a melting beyond the original α -Si layer [96],[110]. The melt front plateau shown in figure 4.6 (b) signifies the time held before the melting of the c-Si. The melted region resolidifies upon reaching the maximum melt depth.

Figure 4.7 shows the dopant retention of the c-Si and the PAI-Si samples as a function of the corresponding maximum melt depth. The dopant retention represents the ratio of the remaining dose extracted from the annealed SIMS profiles to the as-implanted

SIMS profiles (see figure 4.2). The corresponding maximum melt depth is taken from figure 4.5. The dopant retention shows an exponential relation with the melt depth for both c-Si and PAI-Si samples and reaches a constant value of about 73% for melting beyond 80 nm. A $\pm 5\%$ experimental tolerance in the dopant retention is due to a nonuniform native oxide layer on the surface of the c-Si and the PAI-Si samples. The dopant loss mechanism includes boron out-diffusion into the annealing ambience and boron segregation at the native oxide/Si substrate (o/s) interface [113]–[116]. In the presence of a thin native oxide layer on the surface of the Si substrate, boron segregation at the o/s interface becomes more critical. (Further discussion on the components of laser induced dopant loss will be presented in section 4.3.2.) The dopant retention results shown in figure 4.7 can be generalized by an empirical equation [117],

$$DR = DR_{\min} (1 + re^{-\frac{x}{x_o}}) \alpha + (1 - \alpha) \quad (\text{Equation 4.1})$$

where x is the laser induced melt depth, $DR_{\min}$ represents a minimum retention at a deep melting (i.e. 73% in our study), r is the exponential constant, x_o is the depth constant, and α is the percentage of the dopant profile encompassed in the melted region, which can be taken as 1 for a melting that covers greater than 99% of the dopant profile. The values of $DR_{\min}$, r and x_o are determined by the conditions of the silicon surface (i.e. native-oxide capped in our study) and as-implanted dopant profile. The exponential relation of dopant retention with laser induced melt depth is attributed to the redistribution of dopant atoms upon the melting of the samples. Upon the melting of the samples, boron atoms are free to diffuse downwards (into the substrate bulk) and upwards (to the substrate surface). At the surface of the samples, boron atoms could diffuse out of the substrate into the ambience and/or be trapped at the o/s interface. A deeper melting into the samples

redistributes the boron atoms to a lower concentration. The trapped boron atoms, on the other hand, exist in a high concentration that prevents further dopant loss at the interface. This creates an exponential dopant retention relation with melt depth that remains constant at deep melting.

Figure 4.7. Dopant retention (or loss) as a function of melt depth in crystalline and preamorphized Si substrates extracted from the SIMS profiles in figure 4.2 and from SIMS profiles described in figure 4.4.

From figure 4.8, it is observed that the sheet resistances in both c-Si and PAI-Si samples decrease with increasing laser fluence, indicating an increase in the dopant activation [22]. The sheet resistances in both c-Si and PAI-Si samples reaches a minimum value of $150 \, \Omega/\square$ for fluences greater than $0.73 \, \text{J}/\text{cm}^2$. This implies that dopant activation has reached a maximum dopant activation upon deep melting of both c-Si and PAI-Si samples. It is important to note that non-melt LA at $0.5 \, \text{J}/\text{cm}^2$ has induced

moderate activation corresponding to a moderate value of the sheet resistance, as shown in figure 4.8. In order to obtain the activated dopant concentration N , we assume a hole mobility described by Arora *et al.* [118] The model is described as

$$\mu_p = 54.3 + \frac{406.9}{1 + \left(\frac{N}{2.35 \times 10^{17}}\right)^{0.88}} \quad (\text{Equation 4.2})$$

and the corresponding sheet resistance R_s is

$$R_s = \frac{1}{q \int N \mu_p dx} \quad (\text{Equation 4.3})$$

where μ_p is the hole mobility, x is the vertical dimension of the junction and q is the electronic charge. Assuming an active dopant profile that follows the chemical SIMS profile (see figure 4.2) and using the sheet resistance results shown in figure 4.8, the amount of the active dopant of the laser annealed c-Si and PAI-Si samples is calculated and shown in figure 4.9. The dopant activation depicted in figure 4.9 is defined as the percentage of the calculated active dose to the as-implanted dose. It is observed that the dopant activation in both c-Si and PAI-Si samples increases linearly with increasing laser fluences. Higher dopant activation in the PAI-Si sample upon LA at 0.5 and 0.6 J/cm² is attributed to a deeper melting (i.e. both defined by the original α -Si layer) as compared with that in the c-Si sample. For laser fluences greater than 0.73 J/cm², the dopant activation in both c-Si and PAI-Si samples sees a maximum activation at about 75%. When compared with the dopant retention for laser fluences greater than 0.73 J/cm², shown in figure 4.7, the dopant activation corresponds to a 100% activation of the retained dopant in both samples.

Figure 4.8. Sheet resistance of crystalline and preamorphized Si substrate subjected to single-pulse LA at various laser fluences.

Figure 4.9. Dopant activation as a function of laser fluence in crystalline and preamorphized Si substrates. Dopant activation obtained using Hall measurement is also included. The as-implanted B⁺ dose is 1×10¹⁵ cm⁻².

Hall dopant activation included in figure 4.9 is defined as the percentage of the active dopant dose measured by a Hall measurement to the as-implanted dopant dose. The Hall dopant activation for both c-Si and PAI-Si samples shows a good agreement with the calculated values. Equation 4.2 gives a fair accuracy for dopant concentration up to 10^{20} cm^{-3} [118], which falls in the same range of active dopant concentration in figure 4.9. The dopant activation measured by Hall measurement also validates calculated active dopant concentration. However, The Hall dopant activation of the c-Si sample subjected to the non-melt LA at 0.5 J/cm^2 shows a discrepancy of 15% as compared with the calculated value. This is due to a possible under-estimation from the calculation assumption that the active dopant profile follows the chemical profile. Since the 100% dopant activation occurs at high fluences that induce deep melting, a less than the maximum dopant activation is resulted for lower fluence LA, which is typically used for ultra-shallow junction formation. The following section will discuss the use of multiple-pulse LA for further activation of the ultra-shallow junction.

4.3 Multiple-Pulse LA on Boron Implanted Crystalline Si Substrate

4.3.1 Dopant redistribution and activation upon multiple-pulse LA

Figures 4.10 (a) and (b) show the SIMS profiles of a 0.5 keV , $1 \times 10^{15} \text{ cm}^{-2}$ boron-implanted c-Si sample upon shallow-melt LA at 0.6 J/cm^2 and non-melt LA at 0.5 J/cm^2 , respectively, with various number of laser pulses. It is observed that subsequent laser irradiations do not significantly alter the junction depth and junction abruptness annealed by the initial laser irradiations [117], as shown in figure 4.10 and figure 4.11. As shown in figure 4.12, LA at 0.6 and 0.5 J/cm^2 raises the (simulated) surface temperature of the c-

Si sample above 1000 °C in a duration of less than few tens of nanoseconds (see lower abscissa of figure 4.12). The surface temperature eventually cools down to a room temperature at a time greater than 1 μ s (see upper abscissa of figure 4.12). This implies that the initial laser irradiation would not have any interaction with the subsequent laser irradiations in a 1 Hz repetition rate used in this study. Each of the laser irradiation can, thus, be considered as a separate and independent entity. Since each of the laser irradiation at 0.6 J/cm² induces a similar melt depth of 23 nm, the junction depth and junction abruptness remain unchanged upon multiple-pulse LA, as shown in figure 4.12. Non-melt LA at 0.5 J/cm², on the other hand, causes no significant melting to the c-Si sample and induces a negligible diffusion of the as-implanted profile (see figures 4.10 (b) and 4.11). Therefore, the junction depth and junction abruptness also remain unchanged with increasing laser pulses (see figure 4.12).

Figure 4.10. SIMS profiles of boron in crystalline Si substrate subjected to (a) shallow-melt LA at 0.6 J/cm² and (b) non-melt LA at 0.5 J/cm² with various number of laser pulses.

Figure 4.11. Junction depth, determined at $1 \times 10^{18} \text{ cm}^{-3}$, and junction abruptness taken from the SIMS profiles subjected LA at various laser fluences and number of laser pulses shown in figure 4.10. The junction depth and the junction abruptness for the c-Si sample annealed 0.63 J/cm^2 are also included for comparison.

Figure 4.12 shows the (SLIM) simulated surface temperature of c-Si sample upon laser irradiation at 0.6 and 0.5 J/cm^2 . Upon laser irradiation at 0.6 J/cm^2 , the deposited energy raises the surface temperature of the substrate until it reaches the melting temperature of the substrate. The flat portion of the surface temperature profile represents the melting of the substrate, which is defined by the latent heat of melting of the c-Si sample, and the width of the flat portion represents the duration during which the c-Si sample remains in liquid phase [96]. Laser irradiation at 0.5 J/cm^2 , on the other hand, does not generate a sufficiently high temperature that causes melting of the c-Si sample. Nevertheless laser irradiation at 0.5 J/cm^2 induces a surface temperature greater than

1000 °C in few tens of nanoseconds. This is sufficient in inducing dopant activation to the c-Si sample, which corresponds to the moderate sheet resistance value shown in figure 4.13.

Figure 4.12. Simulated surface temperature of c-Si sample upon laser irradiation at 0.6 and 0.5 J/cm² with two different abscissa time scales (i.e. upper μs abscissa and lower ns abscissa).

Figure 4.13 shows that the sheet resistances laser annealed at 0.5 and 0.6 J/cm² decrease with increasing laser pulses. The sheet resistance annealed at 0.63 J/cm², which shows similar decrease with increasing laser pulses, is also included for comparison. This corresponds to an increase in dopant activation upon multiple-pulse LA. Using the same calculation method described in section 4.2, the calculated dopant activation induced by the multiple-pulse LA is summarized in figure 4.14. The figure shows an enhanced dopant activation with increasing laser pulses [119]. As LA at 0.5, 0.6 and 0.63 J/cm²

have not reached a maximum dopant activation upon the initial laser pulse (see figure 4.9), subsequent laser pulses can still promote further incorporation of the boron atoms into the substitutional sites, leading to a higher dopant activation.

Figure 4.13. Sheet resistance as a function of number of laser pulses at various laser fluences.

Figure 4.14. Calculated dopant activation as a function of number of laser pulses at various laser fluences.

4.3.2 Dopant loss upon multiple-pulse LA

Despite that the junction depth and junction abruptness remain unchanged upon multiple-pulse LA, the concentration of the box-like profile, induced by LA at 0.6 J/cm² in figure 4.10 (a), reduces slightly upon multiple-pulse LA. This indicates a dopant loss upon increasing the number of laser pulse. As shown in figure 4.15, the dopant retention decreases with increasing number of pulses (i.e. labeled as native 0.6 J/cm²). The dopant retention is extracted using the same method described in section 4.2 and from the SIMS profiles in figure 4.10 (a). It is observed that a significant portion of the dopant loss occurs at the initial laser irradiation [117] (i.e. 16% at first pulse). The subsequent 99 laser pulses cause only a 12% dopant loss. It is important to note that the dopant retention has an exponential relation with the laser induced melt depth and is attributed to the redistribution of dopant atoms upon the melting of the samples (refer to discussion in

section 4.2). Upon the laser induced melting of the substrate, boron atoms could diffuse out of the substrate into the ambience and/or be trapped at the o/s interface.

Figure 4.15. Dopant retention (or loss) as a function of number of laser pulses with various surface conditions. Label “native” represents the sample with native oxide, “SiO₂” represents the sample capped with 65-nm oxide while “HF” represents the sample pre-treated with diluted HF prior to LA.

In order to identify the contribution of the two aforementioned dopant loss mechanisms, two pre-LA surface treatment methods were conducted. The first method involves the removal of the surface native oxide by diluted HF prior to the multiple-LA at 0.6 J/cm². The laser annealed SIMS profiles is shown in figure 4.16. It is observed that LA at 0.6 J/cm² results in a similar melt depth in the diluted HF-treated sample as compared with that of the sample with the native oxide (see figure 4.10 (a)). As silicon dioxide is non-absorbing at 248 nm wavelength, the thin native oxide of thickness about 2 ~ 3 nm does not alter the threshold of melting in the c-Si sample. As shown in figure

4.15, the diluted HF-treated sample incurs a large dopant loss of about 34% upon the first pulse of LA at 0.6 J/cm^2 . Despite having a similar melt depth, the diluted HF-treated sample has a relative free surface for the dopant to out-diffuse from the substrate. Less dopant loss upon subsequent laser irradiations in the HF-treated sample is attributed to a possible native oxide formation upon the initial LA in the normal ambience. As LA induces a substantially high temperature greater than $1000 \text{ }^\circ\text{C}$ on the c-Si surface, presence of oxygen in the annealing ambience can result in oxide growth [120] that prevents further dopant loss.

Figure 4.16. SIMS profiles of boron in HF dipped c-Si samples (i.e. no native oxide) subjected multiple-pulse LA at 0.6 J/cm^2 . Boron profile of a 0.6 J/cm^2 laser annealed normal c-Si sample (i.e. with native oxide) is also included for comparison.

Figure 4.17. SIMS profiles of boron in 65 nm oxide capped c-Si samples subjected multiple-pulse LA at 0.73 J/cm². Boron profiles of a single-pulse 0.6, 0.61 and 0.73 J/cm² laser annealed normal c-Si sample (i.e. with native oxide) are also included for comparison.

The second method involves capping a 65 nm oxide by plasma enhanced chemical vapor deposition (PECVD) on the c-Si sample prior to a multiple-pulse LA at 0.73 J/cm². Figure 4.17 shows the SIMS profiles of boron in the oxide-capped c-Si sample subjected to multiple-pulse LA at 0.73 J/cm² and in the c-Si sample with native oxide subjected to single-pulse LA at 0.6, 0.61 and 0.73 J/cm². The SIMS analysis on the oxide-capped sample is conducted after a removal of the capped oxide layer by diluted HF. The results show that LA at 0.73 J/cm² induced a significantly shallower melting in the oxide-capped c-Si sample as compared with that in the native-oxide c-Si sample. This is due to a surface reflectivity change of the c-Si sample and a heat sink effect of the oxide in the presence of the thick oxide on the c-Si sample [106],[117]. Whelan *et al.* has reported

that the surface reflectivity of the c-Si sample varies with different thicknesses of oxide layer [117]. The presence of a thick oxide on the c-Si sample can significantly affect the LA properties. Our study shows that the native-oxide c-Si sample subjected to LA at 0.61 J/cm^2 has a similar melting to that of the oxide-capped c-Si sample subjected to LA at 0.73 J/cm^2 (see figure 4.17).

Since the dopant retention or loss depends on the extent of laser induced melting (see figure 4.7), the comparison of dopant retention between the native-oxide Si sample and the oxide-capped Si sample is made at LA at 0.61 J/cm^2 and 0.73 J/cm^2 , respectively. The respective laser fluence induces a similar melt depth of about 26 nm in the native-oxide and the oxide-capped c-Si samples. From figure 4.15, the native-oxide c-Si sample subjected to multiple-pulse LA at 0.61 J/cm^2 incurs a similar dopant loss characteristic as that in the native-oxide c-Si sample subjected to multiple-pulse LA at 0.6 J/cm^2 . On the other hand, the oxide-capped c-Si sample subjected to multiple-pulse LA at 0.73 J/cm^2 sees a similar dopant loss characteristic upon the initial laser pulses, namely an increased dopant loss from 1 pulse to 10 pulses. The dopant loss in the oxide-capped c-Si sample remains constant when subjected to laser pulses greater than 10 pulses. Since the oxide-capped c-Si sample does not have a free surface, boron atoms can only be trapped at the o/s interface or diffused into the oxide. As the capped oxide layer remains solid during the melting of the c-Si substrate at the melting temperature of 1415°C , low boron diffusion in the oxide and short laser irradiation time would result in negligible amount of boron in the bulk of the capped oxide layer. This indicates that a large amount of boron atoms is trapped near the o/s interface. It is believed that the trapped boron atoms at the o/s interface prevents further segregation of boron atoms towards the interface due to a

large concentration gradient. The results from the native-oxide, HF-treated and oxide-capped c-Si samples indicate that segregation of boron atoms at the o/s interface plays a critical role in dopant loss during laser annealing. It is important to note that non-melt LA incurs negligible dopant loss upon multiple-pulse LA (see figure 4.10 (b)) and the dopant retention is about 100%.

4.3.3 Surface morphological modification upon multiple-pulse LA

Figure 4.18 shows the top-view AFM image of a boron implanted c-Si sample. It is observed that the surface of the c-Si sample remains smooth and featureless upon boron implantation at 0.5 keV and $1 \times 10^{15} \text{ cm}^{-2}$. On the other hand, morphological changes to the surface occur when the c-Si sample is subjected to LA [120],[121], as shown in figure 4.19. Figures 4.19 (a) and (b) show a roughened surface of the c-Si sample upon a 5-pulse and 10-pulse LA, respectively, at 0.5 J/cm^2 . It is also observed that the surface becomes rougher when more laser pulses are applied to the c-Si sample (see figures 4.19 (a), 4.19 (b) and 4.20).

Figure 4.18. Top-view AFM images of a 0.5 keV, $1 \times 10^{15} \text{ cm}^{-2}$ boron-implanted c-Si sample.

Figure 4.19. Top-view AFM images of a c-Si sample subject to LA at (a) 5-pulse 0.5 J/cm^2 , (b) 10-pulse 0.5 J/cm^2 and (c) 10-pulse 0.6 J/cm^2 . (d) The three-dimensional AFM image of the c-Si sample subjected to LA at 10-pulse 0.6 J/cm^2 shown in (c).

As shown in figure 4.20, the surface of the c-Si sample remains relatively smooth when subjected to single-pulse LA at 0.5 J/cm^2 and 0.6 J/cm^2 , corresponding to a small value of the root-mean-square roughness comparable to that of the as-implanted sample. The 5-pulse LA at 0.5 J/cm^2 causes the formation of distinctive “islands” on the c-Si sample surface (see figure 4.19 (a)). When the number of laser pulses increases to 10 pulses, “ridges” is formed on the surface as shown in figure 4.19 (b). As shown in figure 4.19 (c), the 10-pulse LA at 0.6 J/cm^2 creates a similar surface roughening to the c-Si samples. Figure 4.19 (d) shows a three-dimensional AFM image of the c-Si sample

subjected to LA at 10-pulse 0.6 J/cm². The LA at 0.6 J/cm² causes a significantly rougher surface as compared with that at 0.5 J/cm². This is due to an increase in the surface roughness with increasing laser fluence that causes redistribution of material at the surface upon resolidification of the melted region [120].

Figure 4.20. Root-mean-square surface roughness of the c-Si sample subjected to multiple-pulse LA in normal or N₂ ambience.

Lu *et al.* has attributed the laser induced surface roughening to the surface waves that arise from surface roughness and inhomogeneity, freezing of capillary waves and generation of transient periodic heating pattern during laser irradiation [121]. Surface roughening observed in the study conducted by Lu *et al.* involves melting of the c-Si sample, which is similar to our study at LA at 0.6 J/cm². As discussed in the previous section, LA at 0.6 J/cm² induces melting to the c-Si sample while LA at 0.5 J/cm² does not induce any melting despite heating up the c-Si sample to a temperature above 1000 °C. Although no melting is induced at 0.5 J/cm², the temperature induced on the surface

of the c-Si sample is sufficiently high to cause redistribution of the material at the surface. This would cause surface morphological changes to the c-Si sample surface.

Figure 4.21. Top-view AFM images of the c-Si sample subject to LA at (a) 5-pulse 0.5 J/cm², (b) 10-pulse 0.5 J/cm² in an N₂ ambience.

As the LA was conducted in a normal ambience, the high temperature induced by the laser irradiation can cause oxide growth due to the presence of oxygen [120]. The surface waves due to surface roughness and inhomogeneity, freezing of capillary waves and generation of transient periodic heating pattern during laser irradiation would cause a nonuniform oxide growth on the c-Si sample surface, leading to surface degradation. Figures 4.21 (a) and (b) show the top-view AFM images of the c-Si sample subject to LA at 5-pulse and 10-pulse LA, respectively, at 0.5 J/cm² in an N₂ ambience. It is observed that a relatively smooth surface is resulted upon LA in the N₂ ambience as compared to that in normal ambience as shown in figure 4.20. This is due to a reduced nonuniform oxide formation on the c-Si sample surface with a decrease in the oxygen partial pressure [120]. However, the laser induced surface roughening is not completely eliminated due to the presence of surface material rearrangement described by Lu *et al.* The drastic increase

in the surface roughness upon high-fluence multiple-pulse LA will cause adverse effects on the electrical characteristics to the MOSFET devices. Leakage current can be significant arising from an uneven distribution of junction depth and non-uniform silicide formation due to the roughened Si surface [2]. Therefore, both laser fluence and number of laser pulses used in multiple-pulse LA should be kept reasonably low to ensure a smooth laser annealed substrate surface.

4.4 Effect of Different Boron Implantation Conditions

Figures 4.22 (a) and (b) show the boron profiles of 0.2, 0.5 and 1.0 keV boron implanted samples subjected to single-pulse LA at 0.63 J/cm^2 and 0.5 J/cm^2 , respectively. The 0.2 keV boron implanted c-Si sample has a dose of $5 \times 10^{14} \text{ cm}^{-2}$ while the 0.5 and 1.0 keV boron implanted c-Si samples have a same dose of $1 \times 10^{15} \text{ cm}^{-2}$. It is observed that the variations in the boron implantations do not change the LA characteristics, resulted in similar melting or non-melting for all the 0.2, 0.5 and 1.0 keV boron implanted samples. From figure 4.23, the LA at 0.63 J/cm^2 creates similar junction depth and junction abruptness in the 0.2, 0.5 and 1.0 keV samples due to a similar melting in all sample (see figure 4.22 (a)). The lower dopant concentration of the 0.2 keV sample as compared to that of the 0.5 and the 1.0 keV samples is due to a lower dose (i.e. $5 \times 10^{14} \text{ cm}^{-2}$) used. The laser induced melting is indifferent to the original as-implantation conditions [117]. Since the induced melting encompasses the entire dopant profiles, the resulted junction profiles are also indifferent to the original as-implantation conditions.

Figure 4.22. SIMS profiles of boron of 0.2, 0.5 and 1.0 keV B⁺ implanted c-Si samples subjected to single-pulse LA at (a) 0.63 J/cm² and (b) 0.5 J/cm².

On the other hand, non-melt LA at 0.5 J/cm² does not induce any melting and the annealed junction profiles are determined by the as-implanted profiles. As a result, the junction depth and the junction abruptness are dependent on the original as-implantation conditions, as shown in figure 4.23. In figure 4.23, The LA at 0.6 J/cm² (detailed results not shown here) shows slight dependence on the original as-implantation conditions. As the LA at 0.6 J/cm² induces a shallower melting (i.e. 23 nm), the tail region of the dopant profiles are not fully encompassed by the shallow melting (see also figure 4.2 (a)), especially true for deeper profile such as in the 1.0 keV sample. As a result, the tail region of the junction profile slightly affects the overall junction profile, depending on the extent of the shallow melting.

Figure 4.23. Junction depth, determined at $1 \times 10^{18} \text{ cm}^{-3}$, and junction abruptness taken from the SIMS profiles in figure 4.22. The junction depth and the junction abruptness created by LA at 0.6 J/cm^2 are also included.

Figure 4.24 shows the sheet resistance of the c-Si samples with various boron implantation energies as function of laser fluence at different laser pulses. All samples show a decrease in the sheet resistance with increasing laser fluence and laser pulse. As the 0.5 keV and the 1.0 keV samples have the same dopant dose, the shallow-melt LA at 0.6 J/cm^2 and 0.63 J/cm^2 create a similar sheet resistance values at 1-, 5- and 10-pulse LA. This is attributed to a similar dopant profile in the 0.5 keV and the 1.0 keV samples induced by the shallow-melt LA, as shown in figure 4.22 (a). On the other hand, the non-melt LA at 0.5 J/cm^2 creates a large spread in the sheet resistance values due to a difference in the annealed dopant profiles that are defined by the as-implanted profiles.

Figure 4.24. Sheet resistance of the c-Si sample with various boron implantation energies as function of laser fluence at different laser pulses.

Using the same calculation method described in section 4.2, the calculated dopant activation of the c-Si samples with different boron implantation energies annealed at various laser fluences and laser pulses is shown in figure 4.25. It is observed that all the 0.2, 0.5 and 1.0 keV samples have a similar dopant activation upon shallow-melt LA at 0.6 J/cm² and 0.63 J/cm². This indicates that the percentage of the incorporation of the dopant atoms into the substitutional sites depends on the melting and resolidification induced by the shallow-melt LA. The dopant activation induced by the non-melt LA at 0.5 J/cm² does not show any significant relation with the original implantation conditions.

Figure 4.25. Calculated dopant activation of the c-Si sample with different boron implantation energies annealed at various of laser fluences and laser pulses.

The results of the sheet resistance as a function of junction depth with comparison to the ITRS roadmap requirement [4] are shown in figure 4.26. Despite that the shallow-melt LA at 0.6 and 0.63 J/cm² creates low sheet resistance, the junction depth only meets the ITRS requirement for a 53 nm physical gate length. This is due to the fact that the shallow-melt LA creates deep junctions upon melting of the c-Si substrate. On the other hand, it is observed that combination of ultra-low energy implantation (i.e. 0.2 keV) and multiple-pulse non-melt LA is suitable for nano-scale gate length technologies, for example 32 nm physical gate length.

Figure 4.26. Relationship between the junction depth of crystalline silicon with various B^+ implantation energies annealed at various of laser fluences and laser pulses. Roadmap requirements for the junction depth and sheet resistance for different physic gate length in ITRS 2004 are also included. Pink label is laser annealed at 0.5 J/cm^2 , green at 0.6 J/cm^2 and blue at 0.63 J/cm^2 .

4.5 Enhanced Dopant Activation in Subamorphized Si Substrate upon Non-melt LA

As described in the previous section that non-melt LA results in a moderate activation to the diffusionless junction. Although multiple-pulse non-melt LA improves the activation, a substantial surface roughening upon subsequent laser pulses limits the application of large number of laser pulses to the sample. In this section, we show a further enhancement in the dopant activation of the diffusionless ultra-shallow junction using a combination of a Si^+ subamorphization implantation (SAI) and non-melt LA. In

this section, the B^+ was implanted at an ultralow energy of 0.5 keV with a dose of $1 \times 10^{15} \text{ cm}^{-2}$ on n-type Si (100) substrate. The Si^+ subamorphization implantation at 10 keV with a dose of $1 \times 10^{14} \text{ cm}^{-2}$ was used to subamorphize the c-Si sample. The Si^+ subamorphization implantation was designed such that it is below the amorphization threshold and creates a spatial distribution of vacancies and interstitials in the substrate [15],[54],[122],[123]. The energy and the dose of the Si^+ subamorphization implantation are not sufficiently high to generate a significant disruption to the lattice structure of the substrate. Despite that the Si^+ subamorphization implantation does not create a continuous α -Si layer, a high concentration of excess vacancies and interstitials (spatially distributed) was generated.

Figure 4.27 shows the SIMS profiles of boron in the c-Si and the SAI-Si samples subjected to LA at 0.50 J/cm^2 and 0.60 J/cm^2 . The boron profile of a 30-nm PAI-Si sample subjected to LA at 0.50 J/cm^2 is also included. As shown in figure 4.1, a 30 nm of α -Si layer was created by the Si^+ preamorphization implantation at 10 keV with a dose of $2 \times 10^{15} \text{ cm}^{-2}$. The as-implanted boron profiles in the SAI-Si and the PAI-Si samples show a reduction in the channeling tail. The 30-nm α -Si layer in the PAI-Si sample effectively blocks the channeling of boron atoms during implantation [124]. Despite that the Si^+ subamorphization implantation in the SAI-Si sample did not create a continuous α -Si layer, the excess vacancies (lying between the surface to about 0.8 of the projected range) create sufficient strain in the lattice, which prevents ion channeling [15]. A similar box-like profile is resulted in the c-Si and the SAI-Si samples in the shallow-melt regime of 0.60 J/cm^2 . This is possible due to the fact that the diffusivity of boron in liquid Si is about eight orders higher than that in solid Si [22],[24]. On the other hand, the non-melt LA at 0.50

J/cm² did not induce any substantial melting [26],[27] and created dopant profiles that follow the as-implanted profiles in both c-Si and SAI-Si samples.

Figure 4.27. SIMS profiles of boron in c-Si and SAI-Si samples subjected to a single-pulse LA at 0.50 J/cm² and at 0.60 J/cm². Boron profile of a 30-nm PAI-Si sample, created by a Si⁺ preamorphization implantation at 10 keV with a dose of 2×10¹⁵ cm⁻², subjected to LA at 0.50 J/cm² is also included.

The PAI-Si sample on the other hand experienced a melting of 30 nm (as seen from the SIMS profile in figure 4.27) due to a lower melting point of the α -Si layer [24]. We have found that a similar threshold of melting at 0.52 J/cm² in the c-Si and the SAI-Si samples despite the fact that the SAI-Si sample contains an additional spatial distribution of excess vacancies and interstitials generated by the Si⁺ subamorphization implantation. Figures 4.28 (a) and (b) show the boron profiles in the c-Si and the SAI-Si samples, respectively, subjected to a non-melt multiple-pulse LA at 0.50 J/cm². The near zero

thermal budget induced by LA causes a negligible boron diffusion even upon multiple-pulse LA, and the annealed profiles follow the as-implanted profiles. The diffusionless profiles in both c-Si and SAI-Si samples show a decrease in the sheet resistance with increasing laser pulses (see figures 4.29 and 4.30). The sheet resistance reaches a minimum after a 20-pulse laser irradiation.

Figure 4.28. SIMS profiles of boron in (a) c-Si and (b) SAI-Si subjected to a multiple-pulse LA in non-melt regime at 0.50 J/cm².

The sheet resistance measurement in both c-Si and PAI-Si samples (depicted in figure 4.29) shows a decrease in the sheet resistance with increasing laser fluence, indicating an increased dopant activation at higher fluence. The sheet resistance reaches a maximum activation corresponding to a sheet resistance value of about 170 $\Omega/\square$ at fluences greater than 0.65 J/cm². It is important to note that fluences greater than 0.55 J/cm² produce a similar activation, corresponding to similar sheet resistance values, in both c-Si and SAI-Si samples. This range of fluences is sufficient to induce a shallow melting to both substrates

that consumes the entire implanted boron profile (see figure 4.27). A liquid phase epitaxy regrowth takes place and creates a c-Si layer from the melted region [25]. On the other hand, the SAI-Si sample shows a lower sheet resistance upon LA in the non-melt regime of about 0.48 J/cm² to 0.5 J/cm² as compared to that of the c-Si sample. The enhanced activation in the SAI-Si sample is attributed to a vacancy-rich region near the surface that are available for dopant incorporation [15],[86] upon non-melt LA.

Figure 4.29. Sheet resistance as a function of laser fluence with an inset showing sheet resistance as a function of laser pulses at a fluence of 0.48 J/cm².

Upon the Si⁺ subamorphization implantation, the forward momentum of the generated interstitials leads to a deeper distribution to that of vacancies. The spatially separated Frenkel pairs introduce an excess vacancy rich region near the surface and an excess interstitials rich region near the implanted tail. TRIM [93] and UT-MARLOWE [94] implantation simulations indicate that the Si⁺ subamorphization implantation generates a projected range of about 18 nm. A distribution of excess vacancies is available from the

surface to about 12 nm. This vacancy-rich region covers greater than 90% of the implanted boron profile. Upon non-melt LA, the vacancy-rich region enhances the incorporation of boron atoms into substitutional sites, leading to an enhanced boron activation seen at 0.48 J/cm² and 0.5 J/cm². The effect of this vacancy-rich region is augmented at reducing laser fluences. However, a larger spreading of the sheet resistance values is seen at low fluence at 0.48 J/cm² due to an incomplete annealing. A higher and more thorough activation can be achieved through multiple-pulse LA, as depicted in the inset of figure 4.29. The decrease in the sheet resistance with increasing laser pulses incurs no additional diffusion in the dopant profiles, as shown in figures 4.28 (a) and (b). The sheet resistance in both c-Si and SAI-Si samples at 0.48 J/cm² reaches a minimum upon 20 pulses of laser irradiations. Dopant activation, defined as the percentage of the active dose to the as-implanted dose measured by a Hall measurement, in the SAI-Si sample (i.e. of 24% activation) shows an enhancement of 10% at single-pulse LA at 0.48 J/cm² as compared to that in the c-Si sample.

Figure 4.30 shows the dopant activation and sheet resistance as a function of laser pulses at 0.5 J/cm², measured by the Hall measurement. A difference in the sheet resistance value at 0.5 J/cm² is seen when compared with the sheet resistance measured by the four-point-probe method shown in figure 4.29. The difference is within 10% of the experimental tolerance due to a square geometry and indium contacts used in the Hall measurement. Similar to the case of 0.48 J/cm², the non-melt LA, 0.5 J/cm² produces a lower sheet resistance in the SAI-Si sample as compared to that of the c-Si sample at all number of pulses. This corresponds to a consistent enhancement of 5% in the dopant activation. The surface vacancy-rich region in the SAI-Si sample enhances the dopant

activation upon LA in the non-melt regime (i.e. at 0.48 J/cm^2 and 0.5 J/cm^2) through the enhanced incorporation of boron atoms into the substitutional sites. This means that SAI technique is capable of producing a higher activation with the low-fluence non-melt LA, which induces a negligible diffusion to the as-implanted profiles.

Figure 4.30. Hall dopant activation and sheet resistance as a function of laser pulses at a fluence of 0.5 J/cm^2 .

Figures 4.31 (a) and (b) shows the junction leakage characteristics of ultrashallow p^+/n junctions in the c-Si and the SAI-Si samples, respectively, formed by single-pulse LA at 0.60 J/cm^2 and multiple-pulse LA at 0.50 J/cm^2 . From figures 4.27 and 4.28, the LA at 0.60 J/cm^2 induces a shallow melting and creates p^+/n junctions of about 33 nm, as defined at a concentration of $1 \times 10^{18} \text{ cm}^{-3}$. The LA at 0.50 J/cm^2 , on the other hand, results in junctions that follow the as-implanted profiles. Both c-Si and SAI-Si samples show a suppression of the leakage current with high fluence and multiple-pulse LA. Accompanied

by a deeper junction due to melting, a higher fluence shows a high removal efficiency of residual defects [125], leading to a lower leakage current. The cumulative effect of multiple-pulse LA [119] also removes deleterious defects at each laser irradiations and creates a lower leakage junction. At 0.6 J/cm^2 , the SAI-Si diode shows a higher voltage dependence in the leakage current, resulting in higher leakage current as compared with that of the c-Si diode. This is due to the presence of excess interstitials not completely annealed by the shallow-melt LA.

Figure 4.31. Junction leakage characteristics of p^+/n junctions in (a) c-Si and (b) SAI-Si subjected to single-pulse LA at 0.60 J/cm^2 and multiple-pulse LA at 0.50 J/cm^2 .

UT-MARLOWE simulation indicates a high concentration of excess interstitials above $1 \times 10^{19} \text{ cm}^{-3}$ distributed up to 100 nm upon the Si^+ subamorphization implantation. A shallow surface melting at 0.6 J/cm^2 leaves a large amount of excess interstitials in the tail region untouched, as shown in figure 4.32 (b). The point defects reside directly adjacent to the junction depletion and create a leaky junction. Despite that the non-melt LA at 0.5

J/cm^2 produces a lower leakage current in the SAI-Si diodes, the reverse current still shows a higher voltage dependence upon 1- and 10-pulse LA. A good reverse current characteristic in the SAI-Si diode is achieved through a 100-pulse non-melt LA. The low leakage characteristic in the SAI-Si diodes upon non-melt LA is due to a possible enhanced annihilation of excess interstitials in the vacancy-rich surface region, as described in figure 4.32 (c). The excess vacancies near the surface not only promote the incorporation of boron atoms but also act as an interstitial sink. Excess interstitials can be readily annihilated through the recombination with the surface excess vacancies.

Figure 4.32. Schematics showing the defect distribution of (a) as-implanted SAI-Si and upon LA in (b) shallow-melt regime and (c) non-melt regime.

4.6 Summary

In summary, we have studied the formation of ultra-shallow p^+/n junctions in silicon substrate upon single- and multiple-pulse LA. Multiple-pulse LA has been shown to be effective in improving the dopant activation without compromising the junction depth and the junction abruptness. The maximum number of laser pulses that can be applied during LA has to be controlled to prevent dopant loss and surface degradation induced by multiple-pulse LA. It is also found that combination of ultra-low energy implantation and multiple-pulse non-melt LA is suitable for nano-scale CMOS technologies. We have also demonstrated the formation of diffusionless ultra-shallow junction in Si^+ subamorphized silicon substrate using non-melt LA. The additional Si^+ subamorphization implantation enhances the dopant activation and improves the electrical characteristics of the junctions in the non-melt regime.

CHAPTER 5

EFFECTS OF RAPID THERMAL ANNEALING IN POST-LASER ANNEALED SILICON

5.1 Introduction

The previous chapter has shown that LA creates a highly activated ultra-shallow junction that could exceed the equilibrium dopant solid solubility limit in the Si substrate. As this supersaturated active dopant concentration exists in a metastable state, the active dopant could deactivate and/or diffuse further upon a post-LA rapid thermal anneal (RTA). In this chapter, we study the thermal stability of laser-activated boron in crystalline, Si^+ and Ge^+ preamorphized Si substrate upon post-laser RTA. We also study the importance of a low temperature anneal in the removal of crystalline defects resulted from pulsed LA of preamorphized ultrashallow junction.

5.2 Boron Diffusion and Deactivation in Crystalline Si Substrate upon Post-LA RTA

B^+ was implanted at 0.5 keV with a dose of $1 \times 10^{15} \text{ cm}^{-2}$ on an n-type Si (100) substrate. LA was carried out in a normal ambience using a 248 nm KrF excimer laser with a pulse duration of 23 ns. Following the LA, the samples were subjected to a 30-s isochronal RTA at temperature between 700 °C and 1000 °C. Figure 5.1 shows the sheet resistance of c-Si samples subjected to various LA conditions followed by the 30-s isochronal RTA. The sheet resistance of the c-Si sample subjected to RTA only is also

included for comparison. The sheet resistance of sample subjected to LA only is represented at 0 °C. Single-pulse and 10-pulse non-melt LA at 0.5 J/cm² result in low sheet resistance values. It is observed that the dopant activation increases with increasing laser pulses, corresponding to a lower sheet resistance with increasing laser pulses. Despite that the LA 0.5 J/cm² induces a high temperature that exceeds 1000 °C (refer to section 4.3), the reaction time of tens of nanoseconds does not causes significant diffusion to the dopant profile, leading to a diffusionless dopant profile (see figure 5.2) [26],[27]. Single-pulse shallow-melt LA at 0.6 J/cm² results in a very low sheet resistance value, which corresponds to a high dopant activation. The LA at 0.6 J/cm² induces a shallow melting to the surface region of the substrate that recrystallizes into a perfect crystalline structure [24],[25]. As the diffusivity of boron in liquid Si is about 8 orders higher than that in solid Si, a box-like dopant profile is resulted upon the LA at 0.6 J/cm² (see figure 5.2). The recrystallization process of the melted Si leads to dopant atoms trapped in the substitutional sites, resulting in a dopant activation that could exceed the dopant solid solubility in the Si substrate.

As shown in figure 5.1, the c-Si sample subjected to the single- and 10-pulse non-melt LA at 0.5 J/cm² incurs an increase in the sheet resistance upon a 30-s isochronal RTA at temperature between 700 °C and 800 °C. A higher temperature of the post-LA RTA (i.e. greater than 850 °C) causes the sheet resistance to decrease. The Hall dopant activation as shown in figure 5.3 (a) indicates that the initial increase in the sheet resistance corresponds to a dopant deactivation upon the post-LA RTA between 700 °C to 800 °C. Further increase in the temperature (i.e. greater than 850 °C) of the post-LA RTA leads to the reactivation of dopant. The Hall dopant activation is defined as the percentage of the Hall

active dose to the as-implanted dose. The c-Si sample received the LA 0.5 J/cm^2 and post-LA RTA incurs a maximum deactivation of about 11% at 850°C . The monotonous increase in the Hall mobility in figure 5.3 (b) is attributed to a reduction in the scattering sites of residual defects and ionized charges with increasing temperature [126],[127]. A higher post-LA RTA temperature removes a higher amount of residual defects left unannealed by the initial LA, reducing the scattering sites that degrade the mobility. On the other hand, the deactivation induced by the post-LA RTA at higher temperature also reduces the ionized impurity scattering. This indicates that the deactivated dopant atoms exist in a neutral structure and do not lead to significant scattering.

Figure 5.1. Sheet resistance of a c-Si sample subjected to various LA conditions and followed by a 30-s isochronal RTA at temperature from 700°C to 1000°C . Prior to the RTA, the c-Si samples were annealed with single-pulse LA at 0.5 J/cm^2 and 0.6 J/cm^2 and 10-pulse (10P) LA at 0.5 J/cm^2 . The sheet resistance of the c-Si sample subjected to RTA only is also included for comparison.

It is also found that significant diffusion in the dopant profile subjected to LA at 0.5 J/cm² occurs for the post-LA RTA temperature greater than 800 °C, as shown in figures 5.2 and 5.4. The diffusion length shown in figure 5.4 is determined at a concentration of $1 \times 10^{18} \text{ cm}^{-3}$. As shown in figure 5.2, the dopant profile subjected to the LA at 0.5 J/cm² followed by the RTA at 800 °C for 30 s sees a noticeable diffusion in the profile tail. The high concentration dopant profile peak, on the other hand, experiences little diffusion. It is also observed that the c-Si sample subjected to the single-pulse LA at 0.5 J/cm² followed by the RTA at 800 °C for 30 s experiences a similar diffusion process to that of the sample subject to RTA only (see figure 5.4). This indicates that the single-pulse LA at 0.5 J/cm² may not be efficient in the removal of lattice damage or defects induced by the implantation [27]. Despite that the LA 0.5 J/cm² induces a high temperature greater than 1000 °C, the extremely short reaction time of tens of nanoseconds does not lead to significant annihilation or evolution of secondary defects [27]. The laser activated dopant atoms are still surrounded by unannealed point defects, small defect clusters and amorphous pockets. The subsequent post-LA RTA provides sufficient thermal energy for dopant diffusion as well as for the formation of extended defects, such as boron-interstitial clusters (BICs). High temperature post-LA RTA greater than 800 °C leads to the dissolution of BICs, a process which releases a large amount of excess interstitials [37]–[39]. This injection of excess interstitials interacts with boron atoms and enhances the diffusion of boron atoms during the post-LA RTA [37]–[39]. It should be noted that the c-Si sample subjected to LA at 0.5 J/cm² followed by the RTA of 1000 °C has a similar sheet resistance with that of the sample subjected to RTA only. This corresponds to a large extent of the dopant diffusion of greater than 70 nm, which involves the laser

activated dopant atoms move out from the substitutional sites. As shown in figure 5.2, a higher efficiency of defect removal upon LA at 0.5 J/cm² occurs at an increasing laser pulses. It is observed that boron atoms in the c-Si sample received the 10-pulse LA 0.5 J/cm² diffuse less as compared to that in the sample received the single-pulse LA. The 10-pulse LA 0.5 J/cm² not only provides higher degree of dopant activation, but also removes a higher amount of the residual defects that assists enhanced diffusion [27],[128].

Figure 5.2. SIMS profiles of boron in the c-Si sample subjected to single-pulse LA at 0.5 J/cm² and 0.6 J/cm² and 10-pulse LA at 0.5 J/cm² (i.e. labeled as 10P 0.5 J/cm²) followed by a RTA at 800 °C for 30 s.

The c-Si sample that received a single-pulse LA at 0.6 J/cm² followed by the 30-s isochronal RTA experiences an initial decrease in the sheet resistance up to a temperature of 800 °C, as shown in figure 5.1. The sheet resistance of the 0.6 J/cm² annealed c-Si sample then increases with increasing RTA temperature between 800 °C and 900 °C. Upon

a high temperature RTA of 1000 °C, the sheet resistance of the 0.6 J/cm² annealed c-Si sample decreases to a lower value again. The Hall dopant activation shown in figure 5.3 (a) shows a mirror image of the sheet resistance trend, which has an initial increase in the dopant activation followed by a deactivation between 800 °C and 900 °C. A slight reactivation of the dopant occurs at high temperature of 1000 °C. The dopant incurs a maximum deactivation of about 9% at 900 °C. The Hall mobility shown in figure 5.3 (b), similar to that of the sample annealed at 0.5 J/cm², also shows a monotonous increase with elevating temperature.

Figure 5.3. (a) Hall dopant activation and (b) Hall mobility of the c-Si sample annealed at 0.5 and 0.6 J/cm² as a function of post-LA RTA temperature. The Hall dopant activation is defined as the percentage of the Hall active dose to the as-implanted dose.

Figure 5.4. Boron diffusion length, determined at a concentration of $1 \times 10^{18} \text{ cm}^{-3}$, in the c-Si sample subjected to a single-pulse LA at 0.5 and 0.6 J/cm^2 , followed by a 30-s isochronal RTA at temperature from 700°C to 1000°C . The boron diffusion length of the c-Si sample subjected to RTA only is also included in the figure.

Figures 5.2, 5.4 and 5.5 indicate that the c-Si sample subjected to the LA at 0.6 J/cm^2 experiences a significant diffusion at the RTA temperature greater than 800°C . This boron diffusion upon the post-LA RTA is attributed to a high concentration of unannealed excess interstitials beyond the maximum melt induced by the LA at 0.6 J/cm^2 . As the shallow-melt LA at 0.6 J/cm^2 involves a surface melting of the substrate, which recrystallizes into perfect crystalline structure using the underneath solid c-Si as a recrystallization template. The recrystallized region is believed to be free of residual defects resulted from the implantation [25]. As shown in figure 5.2, this recrystallized layer is about 26 nm (i.e. the maximum melted depth). This shallow melting of 26 nm,

however, is insufficient to consume the entire boron profile, leaving the excess interstitials in the boron tail unperturbed by the LA. Assuming the contribution of excess interstitials by the implanted boron atoms follows the “+1” model [54],[55], the concentration of the unannealed excess interstitials is slightly below $1 \times 10^{18} \text{ cm}^{-3}$. This is a considerably high amount of excess interstitials that could cause significant enhanced diffusion and dopant deactivation.

Figure 5.5. SIMS profiles of boron in c-Si samples subjected to a single-pulse LA at 0.6 J/cm^2 and followed by a 30-s isochronal RTA at 850, 900 and 1000 °C.

As shown in figure 5.4, it is found that the diffusion length of boron in the c-Si sample subjected to LA at 0.6 J/cm^2 followed by the post-LA RTA is always larger as compared to that of the sample subjected to the LA at 0.5 J/cm^2 followed by post-LA RTA and the sample subjected to RTA only [129]. Figure 5.4 also indicates a slight difference in the diffusion mechanism (i.e. corresponds to the different slope of the lines)

among the c-Si sample subjected to the LA at 0.6 J/cm^2 followed by the post-LA RTA, the sample subjected to the LA at 0.5 J/cm^2 followed by the post-LA RTA and the sample subjected to RTA only. The diffusion in the sample subjected to the LA at 0.6 J/cm^2 followed by the post-LA RTA is dominated by a broadening of the dopant profile shoulder. In the two latter cases (i.e. the sample subjected to the LA at 0.5 J/cm^2 followed by the post-LA RTA and the sample subjected to RTA only), the diffusion of the dopant tail is always seen. This difference may be attributed to a difference in the distribution of excess interstitials. The non-melt LA at 0.5 J/cm^2 generates a high surface temperature greater than $1000 \text{ }^\circ\text{C}$ in an extremely short reaction time of tens of nanoseconds. The laser irradiation induces dopant activation through dopant trapping in the substitutional sites. The defect population, on the other hand, also receives minimum modification during the short laser irradiation. As shown in figure 5.4, the sample subjected to single-pulse LA at 0.5 J/cm^2 followed by the post-LA RTA experiences a similar diffusion process as compared to the sample subjected to RTA only. This indicates that the initial single-pulse LA at 0.5 J/cm^2 does not significantly modify the defect population. It is also observed the diffusion in the sample subjected to the LA at 0.5 J/cm^2 undergoes a BIC formation, depicted as the immobile dopant peak [39], and further diffusion occurs during the later stage of post-LA RTA. The shallow-melt LA at 0.6 J/cm^2 induces a surface melting of about 26 nm . The recrystallization of the melted region leads to a lower concentration of excess interstitials adjacent to the laser annealed box-like profile. As shown in figure 5.5, this lower concentration of excess interstitials does not cause any noticeable immobile boron peak upon the post-LA RTA. This indicates that the lower concentration of excess interstitials readily interacts with the adjacent boron atoms, leading to an enhanced diffusion.

Furthermore, the LA at 0.6 J/cm^2 produces box-like dopant profile with a concentration $4.5 \times 10^{20} \text{ cm}^{-3}$, which is significantly higher than that of the equilibrium boron solid solubility in Si substrate [88],[89]. Boron-enhanced diffusion (BED) is also capable of creating a deep diffusion profile upon the RTA [16],[130]. Mannino *et al.* reported an injection of excess vacancies at the maximum melt depth upon laser induced melting and resolidification [131]. Despite that the liquid phase epitaxial resolidification of the melted region of the Si substrate generates a virtually perfect crystal structure in the resolidified region, the process creates a high content of excess vacancies at the maximum melt depth. The excess vacancies can act as a sink for the excess interstitials, leading to an injection of the excess interstitials adjacent to the laser formed dopant profile. This injection of the excess interstitials is also capable of causing a significant diffusion in the dopant profile.

5.3 Boron Stability in Si^+ Preamorphized Si Substrate upon Post-LA RTA

In this part of the study, n-type Si (100) wafers were preamorphized using a Si^+ implantation at 10 keV with a dose of $2 \times 10^{15} \text{ cm}^{-2}$. As shown in section 4.2, the XTEM analysis reveals that the Si^+ implantation creates a continuous amorphous layer of about 30 nm. Following the Si^+ implantation, B^+ was implanted at 0.5 keV with a dose of $1 \times 10^{15} \text{ cm}^{-2}$. LA was carried out in a normal ambience using a 248 nm KrF excimer laser with a pulse duration of 23 ns. The samples were then subjected to a 30-s isochronal RTA at temperature between 700 °C and 1000 °C.

Figure 5.6. SIMS profiles of boron in Si^+ PAI samples subjected to a single-pulse LA at (a) 0.50 J/cm^2 and (b) 0.65 J/cm^2 followed by a RTA at 800°C and 850°C for 30 s.

Figure 5.6 (a) shows the SIMS profiles of boron in the Si^+ PAI sample subjected to a single-pulse LA at 0.50 J/cm^2 followed by a RTA at 800°C and 850°C for 30 s. It is observed that the LA produces a box-like profile defined by the amorphous Si ($\alpha\text{-Si}$)

layer, i.e. 30 nm. This is due to the fact that the melting temperature of α -Si is about 200 °C lower than that of c-Si [22],[24],[110]. As a result, the LA melts the α -Si layer without melting the underlying c-Si substrate. Since the boron diffusivity in liquid Si is about eight orders higher than that in solid Si, boron atoms diffuse and redistribute uniformly in the melted Si, leading to the formation of the abrupt junction [22],[24].

As shown in figure 5.6 (a), an appreciable diffusion in the profile tail beyond the original amorphous/crystalline (a/c) interface occurs when the 0.50 J/cm² laser annealed Si⁺ PAI sample is subjected to 800 °C and 850 °C for 30 s. The profile tail experiences a diffusion length (at a concentration of 1×10^{18} cm⁻³) of about 4.4 nm and 15.7 nm upon the post-LA RTA at 800 °C and 850 °C, respectively. These diffusion lengths are significantly greater than that predicted by Fair's intrinsic diffusivity of boron [55],[61]. The source for this anomalous diffusion is the supersaturation of the excess interstitials in the end-of-range (EOR) region unannealed by the LA [55]. Since the LA at 0.5 J/cm² involves the melting of the α -Si layer, which recrystallizes into a perfect crystalline structure using the underneath solid c-Si as a recrystallization template. The recrystallized region is free of residual defects induced by the implantation. Despite that the LA induces a high temperature of greater than 1000 °C that only melts the α -Si layer, the reaction time is about tens of nanoseconds. As a result, a significant portion of residual defects (e.g. excess interstitials, amorphous pockets, and defect clusters) is left unannealed in the region beyond the original a/c interface [55]. Extended defects such as {311} defects, EOR loops and BICs are expected to be formed upon the initial stage of the annealing process. Upon further annealing, some of the extended defects dissolve and release

excess interstitials, leading to a transient enhanced diffusion that is significantly greater than that predicted by the intrinsic diffusivity.

Figure 5.7. Boron diffusion length, determined at a concentration of $1 \times 10^{18} \text{ cm}^{-3}$, in the Si^+ PAI Si sample subjected to a single-pulse LA at 0.50 and 0.65 J/cm^2 followed by a 30-s isochronal RTA at temperature from 700 °C to 1000 °C.

The LA at 0.65 J/cm^2 causes a deeper melting of 38 nm, beyond the original a/c interface, as shown in figure 5.6 (b). This leads to an over-melting of the original α -Si that consumes 8 nm of the EOR region. The 0.65 J/cm^2 laser annealed sample incurs a different boron diffusion mechanism from that of the 0.50 J/cm^2 laser annealed sample upon the post-LA RTA at 800 °C and 850 °C. The RTA, in this case, leads to a boron diffusion that broadens the dopant profile tail. Figure 5.7 shows the boron diffusion length, determined at a concentration of $1 \times 10^{18} \text{ cm}^{-3}$, in the Si^+ PAI Si sample subjected to the single-pulse LA at 0.50 and 0.65 J/cm^2 followed by the 30-s isochronal RTA at

temperature from 700 °C to 1000 °C. Figure 5.7 clearly indicates that the 0.50 and 0.65 J/cm² laser annealed sample underwent a different diffusion mechanism upon the post-LA RTA. This difference in the boron diffusion is attributed to the different amount of excess interstitials contributed from the EOR region upon the post-LA RTA.

Figure 5.8. Normalized interstitial dose in the EOR region of the Si⁺ PAI sample as a function of laser induced melt depth. The interstitial dose was simulated using a UT-MARLOWE simulation of a Si⁺ implantation at 10 keV with a dose of 2×10¹⁵ cm⁻² that generates an α-Si layer of 30 nm. The interstitial dose was then normalized with respect to the dose at the original a/c interface.

Figure 5.8 shows the normalized interstitial dose in the EOR region of the Si⁺ PAI sample as a function of the laser induced melt depth. The interstitial dose was simulated using a UT-MARLOWE simulation of the Si⁺ implantation at 10 keV with a dose of 2×10¹⁵ cm⁻² that generates a α-Si layer of 30 nm. The interstitial dose was then

normalized with respect to the dose at the original a/c interface. Assuming that the initial LA does not alter the defect distribution beyond the liquid/solid (l/s) interface, a melting of 38 nm results in a 43% reduction in the excess interstitial dose. This indicates that a high amount of the excess interstitials would still remain in the EOR region upon the LA at 0.65 J/cm². The high amount of excess interstitials released from the EOR region causes a considerable boron diffusion upon the post-LA RTA.

Figure 5.9. SIMS profiles of boron in Si⁺ PAI samples subjected to a single-pulse LA at 0.50 J/cm² and 0.65 J/cm² followed by a RTA at 900 °C and 1000 °C for 30 s.

In both 0.50 J/cm² and 0.65 J/cm² laser annealed samples, the initial stage of the post-LA RTA produces boron-rich clusters in the region of high boron concentration and interstitial clusters in the EOR region [38],[39]. Subsequent enhanced diffusion occurs when the EOR clusters ripen and dissolve, releasing a large amount of excess interstitials. Figure 5.7 indicates that the reduction in the amount of excess interstitials in the EOR

region due to over-melting at 0.65 J/cm^2 enhances the ripening and dissolution of EOR defects. This results in the boron diffusion that broadens the 0.65 J/cm^2 laser annealed dopant profile. On the other hand, the post-LA RTA results in a smaller extent of the boron diffusion in the 0.5 J/cm^2 laser annealed sample because of the presence of a larger amount of immobile BICs. It is expected that the formation of BICs driven by the unannealed excess interstitials in the EOR region causes a high dopant deactivation in the 0.5 J/cm^2 laser annealed sample [38],[39]. (The formation of BICs and dopant deactivation will be discussed in the following section)

Figure 5.10. Schematics showing the formation of BICs in PAI samples (a) subjected to LA melting up to the original α -Si layer (b) – (c) and LA melting beyond the original α -Si layer (d) – (e).

Figure 5.9 shows the SIMS profiles of boron in the Si⁺ PAI sample subjected to a single-pulse LA at 0.50 J/cm² and 0.65 J/cm² followed by a RTA at 900 °C and 1000 °C for 30 s. It is observed that boron atoms diffuse further into the Si substrate upon the high RTA at 900 °C and 1000 °C. The post-LA RTA at 1000 °C produces immobile boron build-ups adjacent to the original I/s interface in both 0.50 J/cm² and 0.65 J/cm² laser annealed samples. The formation of the immobile boron build-ups is attributed to the formation of small BICs [38],[39],[129]. Despite that the liquid-phase epitaxial recrystallization of the melted region induced by the LA at 0.50 J/cm² and 0.65 J/cm², a substantially high content of excess interstitials resides in the original EOR region. Upon the subsequent RTA, a large amount of excess interstitials is injected into the high concentration boron profile, leading to the formation of the immobile and inactive BICs. It is found that an increase in the laser fluence from 0.50 J/cm² to 0.65 J/cm² leads to the decrease in the boron build-up peaks, i.e. $1.73 \times 10^{20} \text{ cm}^{-3}$ and $1.20 \times 10^{20} \text{ cm}^{-3}$ respectively. The LA at 0.65 J/cm² induces an over-melting that consumes the original EOR region as shown in figures 5.6 (b) and 5.9. The over-melting reduces the amount of excess interstitials by about 43%, as shown in figure 5.8, leading to a reduction in the formation of the immobile BICs. The schematics in figure 5.10 show the steps involves in the formation of BICs in the PAI sample subjected to LA melting up to the original α -Si layer and LA melting beyond the original α -Si layer. The driving force for the injection of excess interstitials is a concentration gradient. The corresponding XTEM analysis shown in figure 5.11 indicates no observable boron precipitation or large defect clusters in region where the immobile boron build-ups are formed. This further suggests that the immobile boron build-up is caused by microscopic immobile and inactive BICs.

The formation of the immobile BICs upon the post-LA RTA is expected to cause a deactivation of the originally activated dopant. Figure 5.12 shows the sheet resistance of the Si^+ PAI sample subjected to various LA conditions and followed by a 30-s isochronal RTA at temperature from 700 °C to 1000 °C. Prior to the RTA, the Si^+ PAI samples were annealed with a single-pulse LA at 0.50 J/cm², 0.55 J/cm² and 0.65 J/cm² and 10-pulse LA at 0.5 J/cm². The sheet resistance of the sample subjected to LA only is represented at 0 °C. It can be observed that the LA induces a high dopant activation, which corresponds to the low sheet resistance values. The samples subjected to a single-pulse LA at 0.50 J/cm² and 0.55 J/cm² and 10-pulse LA at 0.5 J/cm² incur an increase in the sheet resistance upon the post-LA RTA between 700 °C and 850 °C. The increase in the sheet resistance reaches a maximum sheet resistance value at 850 °C, after which the sheet resistance decreases with elevating temperature [37],[39],[132].

Figure 5.11. Cross-sectional TEM image of a Si^+ PAI sample subjected to a LA at 0.55 J/cm² and followed by a RTA at 1000 °C for 30 s.

Figure 5.12. Sheet resistance of the Si^+ PAI samples subjected to various LA conditions and followed by a 30-s isochronal RTA at temperature from 700 °C to 1000 °C. Prior to the RTA, the Si^+ PAI sample was annealed with a single-pulse LA at 0.50 J/cm², 0.55 J/cm² and 0.65 J/cm² and 10-pulse (10P) LA at 0.5 J/cm². The sheet resistance of sample subjected to LA only is represented at 0 °C.

The Hall dopant activation shown in figure 5.13 (a) indicates that the initial increase in the sheet resistance of the 0.5 J/cm² laser annealed sample corresponds to a dopant deactivation upon the post-LA RTA between 700 °C to 850 °C. The post-LA RTA produces a maximum deactivation of about 31% at 850 °C in the 0.5 J/cm² laser annealed sample. This dopant deactivation is attributed to the formation of the inactive BICs, resulting from the injection of the excess interstitials from the EOR region. It should be noted that there exists a competing process of BICs formation and dissolution. The former leads to dopant deactivation while the latter causes enhanced boron diffusion, as shown in

figures 5.6 (a) and 5.9. From figure 5.13 (a), further increase in the temperature (i.e. greater than 850 °C) of the post-LA RTA leads to the reactivation of the dopant [37],[39],[132].

The 0.65 J/cm² laser annealed sample, on the other hand, incurs a different dopant deactivation characteristics. The post-LA RTA produces an initial decrease in the sheet resistance up to a temperature of 800 °C, as shown in figure 5.12. The sheet resistance of the 0.65 J/cm² laser annealed sample increases with elevating RTA temperature between 800 °C and 900 °C. The sheet resistance of the 0.65 J/cm² laser annealed sample decreases with the post-LA RTA at 1000 °C. The Hall dopant activation shown in figure 5.13 (a) indicates that the initial decrease in the sheet resistance corresponds to a slight increase in dopant activation while the increase in the sheet resistance between 800 °C and 900 °C is due to a dopant deactivation of 9%. The dopant deactivation characteristics are attributed to the reduced formation of BICs due to a less excess interstitial injection upon the over-melting LA at 0.65 J/cm². The monotonous increases with increasing temperature in the Hall mobility in figure 5.13 (b) indicates that deactivated dopant atoms do not contribute to the scattering sites of ionized charges [126],[127]. The results in figures 5.12 and 5.13 (a) indicate that the over-melting of the original Si⁺ preamorphized leads to less excess interstitials and less BICs. This improves the immunity of the boron deactivation against the post-LA RTA.

Figure 5.13. (a) Hall dopant activation and (b) Hall mobility of the Si⁺ PAI samples annealed at 0.50 and 0.65 J/cm² as a function of post-LA RTA temperature.

5.4 Reduction of Boron Diffusion and Deactivation by Ge⁺ Preamorphization upon Post-LA RTA

B⁺ was implanted on n-type Si (100) wafers at 0.5 keV to a dose of $1 \times 10^{15} \text{ cm}^{-2}$. Prior to the B⁺ implantation, the wafers were preamorphized using either Si⁺ or Ge⁺ implantation. The Si⁺ was implanted at 10 keV to a dose of $2 \times 10^{15} \text{ cm}^{-2}$ to create an amorphous layer of 30 nm while the Ge⁺ was implanted at 20 keV to a dose of $1 \times 10^{15} \text{ cm}^{-2}$ to produce an amorphous layer of 34 nm. The samples were irradiated with a 248 nm KrF excimer laser with a pulse duration of 23 ns. Following the LA, the samples were subjected to a 30-s isochronal RTA at temperature between 700 °C and 1000 °C.

Figure 5.14 (a) shows the SIMS profiles of boron in the Si^+ and the Ge^+ PAI samples subjected to a single-pulse LA at 0.50 J/cm^2 followed by a post-LA RTA at 800°C and 850°C for 30 s. It is observed that the LA at 0.50 J/cm^2 creates box-like dopant profiles that are contained in the original amorphous Si ($\alpha\text{-Si}$) layer, i.e. 30 nm for the Si^+ PAI sample and 34 nm for the Ge^+ PAI sample. Similar mechanism described in sections 4.2 and 5.3 is responsible for the formation of box-like profiles. From figure 5.14 (a), it is observed that the post-LA RTA at 800°C and 850°C produces a significant diffusion in the profile tail beyond the original amorphous/crystalline (a/c) interface of the 0.5 J/cm^2 laser annealed Si^+ and Ge^+ PAI samples. This anomalous diffusion is attributed to the injection of supersaturated excess interstitials from the end-of-range (EOR) region untouched by the laser induced melting. It is believed that the liquid phase epitaxial recrystallization of the melted Si layer produces a defect free recrystallized layer. However, the EOR region untouched by the melting has a supersaturation of excess interstitials. The extremely high concentration of the excess interstitials in this region can cause enhanced dopant diffusion and dopant deactivation upon the post-LA thermal processes [38],[39],[133]. The initial stage of the post-LA RTA produces extended defects such as $\{311\}$ defects, EOR loops and boron-interstitial clusters (BIC). Enhanced dopant diffusion can occur upon the dissolution of these defects, which release the mobile interstitials. (Dopant deactivation induced by excess interstitials will be discussed in the following section)

From figures 5.14 (a) and 5.15, the 0.5 J/cm^2 laser annealed Si^+ PAI sample incurs a diffusion length, at a concentration of $1 \times 10^{18} \text{ cm}^{-3}$, of 4.4 nm and 15.7 nm at 800°C and 850°C , respectively. On the other hand, the 0.5 J/cm^2 laser annealed Ge^+ PAI sample

incurs a diffusion length of 3.7 nm and 4.3 nm upon the post-LA RTA at 800 °C and 850 °C, respectively. The results show that the 0.5 J/cm² laser annealed Ge⁺ PAI sample incurs a less boron diffusion as compared to that of Si⁺ PAI sample upon the post-LA RTA. A same observation is also found in the 0.5 J/cm² laser annealed Ge⁺ PAI sample subjected to a post-LA RTA at 900 °C and 1000 °C, as shown in figure 5.14 (b). From figure 5.15, the 0.5 J/cm² laser annealed Si⁺ and Ge⁺ PAI samples have a similar diffusion characteristic, i.e. signified by a similar slope in both lines. It is also observed from figure 5.15 that the 0.5 J/cm² laser annealed Ge⁺ PAI sample consistently incurs a smaller extent of the boron diffusion as compared to that of the Si⁺ PAI sample. The reduced boron diffusion in the Ge⁺ PAI sample is related to the strain compensation induced by the Ge atoms. Previous work has reported that the Si lattice contract when boron is incorporated into the substitutional sites upon LA [88],[89]. The covalent radius of boron atom is substantially smaller than that of Si atom, resulting in a lattice strain. This lattice strain can enhance boron atoms to move out from the substitutional lattice sites upon thermal processes [134]. Incorporation of atoms of larger covalent radius than that of Si atoms, such as Ge atoms, minimizes the strain induced by the boron atoms. The tetrahedral radii of Si, Ge and boron atoms are 1.18 Å, 1.22 Å and 0.88 Å, respectively [135]. This would reduce the amount of substitutional boron atoms moving out from the substitutional lattice sites upon thermal processes. This results in a reduction in the enhanced diffusion and the dopant deactivation upon the post-LA RTA.

Figure 5.14. SIMS profiles of boron in Si⁺ and Ge⁺ PAI samples subjected to a single-pulse LA at 0.50 J/cm² followed by a 30-s isochronal RTA at (a) 800 °C and 850 °C and (b) 900 °C and 1000 °C.

Figure 5.14 (b) shows the SIMS profiles of boron in the Si⁺ and the Ge⁺ PAI samples subjected to a single-pulse LA at 0.5 J/cm² followed by a 30-s isochronal RTA at 900 °C

and 1000 °C. It is observed that high RTA temperature of 900 °C and 1000 °C induces significant boron diffusion in the Si substrate for both Si⁺ and Ge⁺ PAI samples. The Ge⁺ PAI sample produces a smaller extent of boron diffusion due to the strain compensation between the boron and germanium atoms in the Si lattice. Immobile boron build-ups of concentration of $1.73 \times 10^{20} \text{ cm}^{-3}$ and $1.70 \times 10^{20} \text{ cm}^{-3}$ are found adjacent to the original l/s interface in both Si⁺ and Ge⁺ PAI samples, respectively, upon the post-LA RTA at 1000 °C. The formation of the immobile build-ups is attributed to the formation of small BICs [38],[39],[129]. The LA at 0.50 J/cm^2 induces a melting of the α -Si layer created by the Si⁺ and Ge⁺ implants without melting the underlying c-Si substrate. The melted region recrystallizes into a perfect crystalline structure. However, a large amount of excess interstitials is left unannealed in the EOR region by the LA due to a short laser duration. Upon subsequent RTA, the excess interstitials in the EOR region diffuse into the bulk and the surface of the Si substrate. The excess interstitials diffuse into the high boron concentration region also lead to the formation of the immobile and inactive BICs [38],[39]. The formation of BICs also causes dopant deactivation to the initially activated dopant. The dopant deactivation upon post-LA RTA will be further discussed in the following section.

Figure 5.15. Boron diffusion length, determined at a concentration of $1 \times 10^{18} \text{ cm}^{-3}$, in Si^+ and Ge^+ PAI Si samples subjected to a single-pulse LA at 0.50 and 0.65 J/cm^2 followed by a 30-s isochronal RTA at temperature from 700 °C to 1000 °C.

Figures 5.16 (a) and (b) show the SIMS profiles of boron in the Si^+ and the Ge^+ PAI samples, respectively, subjected to a single-pulse LA at 0.65 J/cm^2 followed by a 30-s isochronal RTA at temperatures between 800 °C and 1000 °C. It is found that the Ge^+ PAI sample consistently produces a smaller extent of boron diffusion as compared to that of the Si^+ PAI sample (see also figure 5.15). This is consistent with the case of the 0.50 J/cm^2 laser annealed Si^+ and the Ge^+ PAI samples, as shown in figures 5.14 (a) and (b). From figure 5.16 (a), it is observed that the Si^+ PAI sample produces an immobile boron build-up of $1.20 \times 10^{20} \text{ cm}^{-3}$ adjacent to the original l/s interface upon the post-LA RTA at 1000 °C. An injection of excess interstitials from the EOR region is responsible for the formation of immobile BICs. Due to a over-melting of the original $\alpha\text{-Si}$ layer that

consumes the EOR region, i.e. 8 nm in this case, the boron build-up concentration is less than that of the 0.50 J/cm^2 laser annealed Si^+ PAI sample (see figure 5.14 (b)). On the other hand, the post-LA RTA at 1000°C does not create any noticeable boron build-up in the Ge^+ PAI sample, as shown in figure 5.16 (b). This indicates that the concentration of BICs upon the post-LA RTA at 1000°C has drastically reduced below the “background” boron concentration of $1.06 \times 10^{20} \text{ cm}^{-3}$.

It is expected that the 20 keV Ge^+ implantation produces a higher excess interstitial concentration in the EOR region as compared to that of the 10 keV Si^+ implantation, as shown in figure 5.17. The LA 0.65 J/cm^2 leads to a over-melting of the original $\alpha\text{-Si}$ layer that consumes 6 nm of the EOR region in the Ge^+ PAI sample (see figure 5.16 (b)). As shown in figure 5.17, this over-melting reduces the excess interstitials of about 23% in the interstitials dose. The amount of excess interstitials in the EOR region of the Ge^+ PAI sample is higher than that of the Si^+ PAI sample. It is also found that the post-LA RTA at 1000°C causes negligible Ge diffusion in the laser annealed Ge^+ PAI sample, as shown in figure 5.18. This is attributed to a low Ge diffusivity in the Si substrate [136]. This also indicates that the Ge atoms do not chemically participate in the formation of the BICs. The previous section has discussed that the strain compensation reduces the boron diffusion by reducing the probability of boron atoms moving out from the substitutional sites. This means the interaction between the boron atoms and excess interstitials is disturbed by the presence of Ge atoms. Despite a larger amount of excess interstitials remained in the EOR region of the Ge^+ PAI sample upon the initial LA, the post-LA RTA produces a smaller extent of dopant diffusion and BIC formation. This implies that

the Ge⁺ PAI sample should have a less dopant deactivation upon the post-LA RTA. This is verified in the following discussion.

Figure 5.16. SIMS profiles of boron in (a) Si⁺ and (b) Ge⁺ PAI sample subjected to a single-pulse LA at 0.65 J/cm² followed by a 30-s isochronal RTA at temperatures between 800 °C and 1000 °C.

Figure 5.17. Normalized interstitial dose in the EOR region of the Ge⁺ and the Si⁺ PAI samples as a function of laser induced melt depth. The interstitial dose was simulated using a UT-MARLOWE simulation of Ge⁺ implantation at 20 keV with a dose of $1 \times 10^{15} \text{ cm}^{-2}$ and of Si⁺ implantation at 10 keV with a dose of $2 \times 10^{15} \text{ cm}^{-2}$. The interstitial dose was then normalized with respect to the dose at the original a/c interface.

Figure 5.19 shows the sheet resistance of the Ge⁺ PAI sample subjected to various LA conditions and followed by a 30-s isochronal RTA at temperature from 700 °C to 1000 °C. Prior to the RTA, the Ge⁺ PAI samples were annealed with a single-pulse LA at 0.50 J/cm², 0.55 J/cm² and 0.65 J/cm² and 10-pulse LA at 0.5 J/cm². The sheet resistance of the sample subjected to LA only is represented at 0 °C. The sheet resistance of the sample received single-pulse LA at 0.50 J/cm² and 0.55 J/cm² and 10-pulse LA at 0.5 J/cm² shows a decrease in the sheet resistance value upon the post-LA RTA at 700 °C. The sheet resistance then increases with elevating temperature between 700 °C and 850 °C,

followed by a decrease between 900 °C and 1000 °C. The Hall dopant activation in figure 5.20 (a) indicates a mirror image of the sheet resistance trend of the 0.50 J/cm² laser annealed sample. Figure 5.20 (a) shows that the initial decrease in the sheet resistance at 700 °C corresponds to an increase in the dopant activation while the increase in the sheet resistance between 700 °C and 850 °C corresponds to a dopant deactivation. The high temperature RTA between 900 °C and 1000 °C causes a dopant reactivation. The maximum dopant deactivation of 17% occurs at 850 °C for the 0.50 J/cm² laser annealed Ge⁺ PAI sample. This value is significantly lower than that of the 0.50 J/cm² laser annealed Si⁺ PAI sample, i.e. 31% (see figure 5.13 (a)). This agrees well with the hypothesis that the strain compensation induced by the Ge atoms hinders the formation of inactive BICs.

Figure 5.18. SIMS profiles of germanium in the Ge⁺ PAI sample subjected to LA at 0.5 J/cm² only, LA at 0.5 J/cm² with RTA and RTA only. The RTA was conducted at 1000 °C for 30 s.

Figure 5.19. Sheet resistance of Ge⁺ PAI samples subjected to various LA conditions and followed by a 30-s isochronal RTA at temperature from 700 °C to 1000 °C. Prior to the RTA, the Ge⁺ PAI sample was annealed with a single-pulse LA at 0.50 J/cm², 0.55 J/cm² and 0.65 J/cm² and 10-pulse (10P) LA at 0.5 J/cm². The sheet resistance of sample subjected to LA only is represented at 0 °C.

The post-LA RTA produces an initial decrease in the sheet resistance of the 0.65 J/cm² laser annealed Ge⁺ PAI sample up to a temperature of 800 °C, as shown in figure 5.19. The sheet resistance of the 0.65 J/cm² laser annealed sample increases with elevating RTA temperature between 800 °C and 900 °C. Then the sheet resistance of the 0.65 J/cm² laser annealed sample decreases with the post-LA RTA at 1000 °C. The Hall dopant activation shown in figure 5.20 (a) indicates that the initial decrease in the sheet resistance corresponds to a slight increase in the dopant activation while the increase in the sheet resistance between 800 °C and 900 °C is due to a dopant deactivation of 9%. The reduced

dopant deactivation of the 0.65 J/cm² laser annealed sample as compared to that of the 0.50 J/cm² laser annealed sample is attributed to the reduced formation of BICs due to less excess interstitials injection upon the over-melting LA at 0.65 J/cm². Figure 5.20 (b) indicates a higher Hall mobility at higher fluences due to annihilation of scattering sites at high fluence LA. The monotonous increase in the Hall mobility with increasing temperature also suggests that deactivated dopant atoms exit in a neutral structure and do not contribute to the scattering sites of ionized charges [126],[127]. Furthermore, a higher post-LA RTA temperature removes a higher amount of residual defects left unannealed by the initial LA. This reduces the scattering sites that degrade the Hall mobility. Table 5.1 summarizes the thermal stability of the boron activation and diffusion for various PAI and LA schemes against the post-LA RTA. The results indicate that Ge⁺ PAI improves the immunity of boron diffusion and deactivation against the post-LA RTA.

Figure 5.20. (a) Hall dopant activation and (b) Hall mobility of the Ge⁺ PAI sample annealed at 0.50 and 0.65 J/cm² as a function of post-LA RTA temperature.

Table 5.1. Thermal stability of boron activation and diffusion for various PAI and LA schemes against post-LA RTA.

PAI Scheme	LA Technique	Dopant Deactivation*	Dopant Diffusion
c-Si	Non-melt	11% deactivation at 850 °C	Reduced diffusion with multiple-pulse LA
	Shallow-melt	9% deactivation at 900 °C	Enhanced diffusion
Si ⁺ PAI	Melting up to PAI	31% deactivation at 850 °C	Enhanced diffusion
	Over-melting of PAI	9% deactivation at 900 °C	Enhanced diffusion
Ge ⁺ PAI	Melting up to PAI	17% deactivation at 850 °C	Reduced diffusion as compared with Si ⁺ PAI
	Over-melting of PAI	9% deactivation at 900 °C	Reduced diffusion as compared with Si ⁺ PAI

* Dopant deactivation is defined as the difference between the highest achieved dopant activation and the dopant activation at the specified RTA temperature.

5.5 Role of Low Temperature RTA in Post Laser Annealed Preamorphized pMOSFET

In this part of the study, B⁺ was implanted at 0.5 keV with a dose of 1×10¹⁵ cm⁻² on n-type Si (100) substrate. Prior to the B⁺ implantation, some of the wafers were preamorphized using Si⁺ implantation at 10 keV with a dose of 2×10¹⁵ cm⁻². As shown in section 4.2, the XTEM analysis indicates that a continuous amorphous layer of about 30 nm was created by the Si⁺ implantation. pMOSFET with a thermally grown 6.5 nm gate oxide was also fabricated for the electrical study. After gate stack patterning, the source/drain region was preamorphized using a Si⁺ implantation at 10 keV with a dose of 3×10¹⁵ cm⁻², creating an amorphous layers of 35 nm. B⁺ was implanted at 2.0 keV with a dose of 3×10¹⁵ cm⁻². The 2.0-keV B⁺ implantation was used to overcome a surface oxide

present during the device integration and to optimize a junction overlap region near the gate/spacer edge. p^+/n diodes were also fabricated using the same pMOSFET implantation processes. Based on a TRIM simulation, both boron profiles in the unpatterned and device samples fall within the Si^+ preamorphized layer.

Figure 5.21. SIMS profiles of boron subjected to LA at 0.6 J/cm^2 with or without a post-LA low temperature RTA at $600 \text{ }^\circ\text{C}$, 60 s. Minimum diffusion near the tail region occurred due to by the additional RTA cycle. The B^+ was implanted at 0.5 keV with a dose of $1 \times 10^{15} \text{ cm}^{-2}$ while the substrate surface preamorphization was created using Si^+ implantation at 10 keV with a dose of $2 \times 10^{15} \text{ cm}^{-2}$.

Figure 5.21 shows the SIMS profiles of boron subjected to LA at 0.6 J/cm^2 with or without a post-LA RTA at $600 \text{ }^\circ\text{C}$, 60 s. A box-like boron profile is contained in the original 30-nm PAI region due to the melting of the $\alpha\text{-Si}$ layer only during LA. It is observed that the $600 \text{ }^\circ\text{C}$ low temperature RTA causes minimum diffusion near the tail

region of the laser annealed boron profile. The diffusion is attributed to the presence of EOR defects left unannealed after LA [55]. A Hall measurement reveals a decrease of the corresponding sheet resistance from 286 $\Omega/\square$ to 166 $\Omega/\square$ when the LA sample was subjected to the post-LA low temperature RTA. This corresponds to an improvement of Hall mobility of about 2 times to 25 cm^2/Vs with negligible dopant deactivation. The enhancement in the mobility is caused by a reduction in the crystalline defects annealed out by the additional low temperature RTA that will be explained in the following discussion.

Figure 5.22. I-V characteristics of p^+/n junction subjected to (a) only LA and (b) LA followed by a RTA. Both single-pulse (1P) and ten-pulse (10P) of laser fluences of 0.5 J/cm^2 and 0.6 J/cm^2 were used for (a) and (b). The post-LA RTA in (b) was conducted at 600 °C, 60 s. The B^+ was implanted at 2.0 keV with a dose of $3 \times 10^{15} \text{ cm}^{-2}$ while the substrate surface preamorphization was created using Si^+ implantation at 10 keV with a dose of $3 \times 10^{15} \text{ cm}^{-2}$.

Figure 5.22 shows the diode characteristics of the ultrashallow p^+/n junctions formed using similar LA conditions to those shown in figure 5.21, i.e. by LA only or LA followed by the RTA. Laser fluence of 0.5 J/cm^2 or 0.6 J/cm^2 was used to completely melt the PAI layer without melting of the underlying substrate. The conditions were established from the XTEM, SIMS analysis and using a one-dimensional heat flow calculation [92],[96]. In figure 5.22 (a), single-pulse LA of fluences of 0.5 and 0.6 J/cm^2 resulted in a high junction leakage current of the same order as the forward current. As depicted in figure 5.23 (a), the high leakage current is due to the presence of crystalline defects, such as stacking faults, microtwins, dislocation and EOR damage, extending throughout the original PAI region that could not be removed by the single-pulse LA [119],[125],[137]. Fast recrystallization of the melted PAI region causes the formation of crystalline defects [119],[125],[137]. Even the recrystallized PAI layer is epitaxial with the Si (100) substrate as shown by the diffraction pattern in the inset in figure 5.23 (a), the crystalline defects lead to the highly leaky junctions shown in figure 5.22 (a). It is also observed that a higher fluence at 0.6 J/cm^2 resulted in a lower leakage current, which confirms a higher defect removal efficiency at high fluence [55],[107]. However, high fluence at 0.6 J/cm^2 is sufficient to cause a melting in c-Si and leads to undesirable deformation of the polycrystalline Si gate [108]. On the other hand, a ten-pulse LA suppresses the leakage current by two orders of magnitude [119]. The corresponding TEM results indicate that the crystalline defects were effectively eliminated by the subsequent laser pulses. Figure 5.22 (b) indicates an improved diode characteristic in both single- and ten-pulse LA following a post-LA low temperature RTA at 600°C for 60 s which alone shows very leaky junction. A suppression of the leakage current of about two orders in the single-pulse LA and one order in the ten-pulse

LA is achieved by annealing out the crystalline defects by the low temperature RTA (see figure 5.23 (b)). This signifies the presence of residual defects upon the highly non-equilibrium and fast LA process, especially for the single-pulse LA. On the other hand, dissolution or removal of the crystalline defects occurs when subjected to the relatively more equilibrium process of low temperature RTA. RTA at 600 °C for 60 s alone does not cause significant activation of boron dopants, and the junction remains leaky. A sheet resistance measurement indicates that at 600 °C RTA, a moderate activation of dopants through solid-phase epitaxial regrowth (SPER) [40],[138] of the preamorphized layer only occurs at a longer annealing duration of at least 180 s.

Figure 5.23. Cross-sectional transmission electron microscopy micrographs of unpatterned PAI samples subjected to a single-pulse LA at (a) 0.5 J/cm² only and (b) 0.5 J/cm² with a post-LA RTA at 600 °C for 60s. The inset in figure 5.23 (a) is the electron diffraction pattern showing epitaxial silicon (001) obtained from the recrystallized region that contains crystalline defects. The implantation conditions are the same as those used in figure 5.21.

Figures 5.24 (a) and (b) show the I_d - V_{ds} and I_d - V_{gs} characteristics, respectively, of a 0.5 μm pMOSFET subjected to a single-pulse LA at 0.5 J/cm^2 with a post-LA RTA at 600 $^\circ\text{C}$ for 60 s and a ten-pulse LA at 0.5 J/cm^2 . The low drive current is due to unsilicided source/drain and gate regions. An annihilation of the crystalline defects in the annealed junctions gives rise to a higher current drive and lower leakage current of 2 orders in the pMOSFET subjected to post-LA low temperature RTA. Similar results were obtained for other pMOSFETs fabricated by LA followed by RTA process (i.e. figure 5.22 (b)). On the other hand, no transistor characteristics can be obtained from the pMOSFETs fabricated by only single-pulse LA (i.e. figure 5.22 (a)). This is due to an excessive junction leakage current resulted from the single-pulse LA. Although a higher fluence (i.e. greater than 0.6 J/cm^2) suppresses the junction leakage current, undesirable effects, such as physical gate degradation [108] and deeper junction depth, are accompanied. Therefore, laser fluence used for MOSFET fabrication should be kept low to prevent these undesirable effects while maintaining a low junction leakage current through annihilation of residual defects. As shown in figures 5.24 (a) and (b), this can be achieved by applying a post-LA low temperature RTA. It should be noted that an optimal laser fluence obtained on a unpatterned sample and on a patterned devices may vary due to the presence of oxide isolation and gate stacks [107],[108],[139]. Due to an isolation oxide thermal confinement effect and heat sink effect, a two-dimensional manifestation of deeper junction in the device structure (as compare to unpatterned sample) and shallower junction at edge of active area of the device structure (as compare to the center of the active area) has to be considered. Nevertheless, the results show that the post-LA low temperature RTA is an

important process and effective in suppressing junction leakage for a wide range of laser fluences and conditions used for MOSFET fabrication.

Figure 5.24. (a) Drain current (I_d) versus drain bias (V_{ds}) and (b) drain current (I_d) versus gate bias (V_{gs}) characteristics of a 0.5 μm pMOSFET subjected to a single-pulse LA at 0.5 J/cm² with a post-LA RTA at 600 °C for 60 s or only to a ten-pulse LA at 0.5 J/cm². The low drive current is due to unsilicided source/drain and gate regions. The implantation conditions are the same as those used in figure 5.22.

5.6 Summary

LA produces dopant activation that exceeds the equilibrium solid solubility of boron in the Si substrate. This high dopant activation is metastable and could be deactivated upon subsequent thermal cycles. It is found that laser activated boron atoms remain stable up to 800 °C, after which substantial deactivation takes place in the Si⁺ and the Ge⁺ preamorphized samples. This deactivation is caused by the presence of excess interstitials in the end-of-range region induced by the implantation that could not be annealed by

melting up to the amorphous/crystalline interface. Melting beyond the amorphous layer suppresses dopant deactivation upon post-LA RTA through a reduction of excess interstitials in the end-of-range region. A reduced dopant deactivation and diffusion upon the post-LA RTA is observed in the Ge^+ PAI sample. A post-LA low temperature RTA of 600 °C has been shown to be critical for the further elimination of the crystalline defects in the laser annealed preamorphized pMOSFET. It has been shown that pMOSFETs with good electrical characteristics can be obtained using LA followed by a low thermal budget RTA.

CHAPTER 6

THERMAL CONFINEMENT OF ADVANCED SEMICONDUCTOR SUBSTRATES DURING LASER ANNEALING

6.1 Introduction

Thermal dissipation takes on a crucial role for LA in advanced semiconductor substrates, such as silicon-on-insulator (SOI) and strained-Si-on-SiGe substrates. SOI and strained-silicon/SiGe substrates comprise a thermally insulating or less conductive layer, which is oxide in the SOI substrate and SiGe layer in the strained-silicon/SiGe substrate. The thermally less conductive layers deprive a good thermal dissipation pathway when subjected to LA. A substantially high background temperature can build up due to slow thermal dissipation upon laser irradiation. This gives rise to a laser process that is significantly different as compared to that in normal bulk silicon substrate. This chapter compares the formation of ultra-shallow p^+/n junctions in bulk silicon, SOI and strained-silicon/SiGe substrates using LA in shallow-melt and non-melt regimes.

6.2 Thermal Confinement in SOI Substrate during LA

B^+ was implanted at 0.5 keV or 1.0 keV with a dose of $5 \times 10^{14} \text{ cm}^{-2}$ on n-type Si (100) and SOI substrates. The SOI wafer used has a topmost silicon (100) layer thickness of 182 nm and a buried silicon dioxide (BOX) layer thickness of 188 nm. Prior to the B^+ implantation, some of wafers were preamorphized using Si^+ implantation at 30 keV with

a dose of $2 \times 10^{15} \text{ cm}^{-2}$. A continuous surface amorphous layer of 77 nm, shown in figure 6.1, was generated by the 30 keV Si^+ implantation. The previous chapter has shown that preamorphization provides a process window in LA of which the junction depth is limited by the depth of the amorphous layer. A process window is provided since the melting temperature of amorphous silicon ($\alpha\text{-Si}$) is about 200 °C lower than that of crystalline silicon (c-Si) [22],[24],[110].

Figure 6.1. Cross-sectional TEM micrograph of a boron implanted silicon substrate that was preamorphized with Si^+ implantation at 30 keV, $2 \times 10^{15} \text{ cm}^{-2}$.

6.2.1 LA of preamorphized SOI and Si substrates

Figures 6.2 (a) and (b) show the SIMS profiles of boron in the preamorphized silicon and the SOI substrates respectively, after single-pulse LA at laser fluences of 0.2, 0.4 and 0.6 J/cm^2 . The laser fluences used are all above the threshold fluence for melting amorphous silicon. It can be seen from figures 6.2 (a) and (b) that all the boron profiles in the SOI substrates have a deeper dopant distribution than the profiles in the silicon substrate. In the case of 0.4 and 0.6 J/cm^2 laser fluences in the silicon substrate (figure 6.2

(a)), due to a thick melted layer, boron could not re-distribute fast enough to form a step-like profile. On the other hand, in figure 6.2 (b), step-like profiles are formed in the SOI substrate for the same fluences used for the silicon substrate.

Figure 6.2. SIMS profiles of boron in 30-keV Si⁺ preamorphized (a) silicon and (b) SOI substrates subjected to single-pulse LA at fluence of 0.2, 0.4 and 0.6 J/cm².

The junction abruptness and junction depth taken at $1 \times 10^{18} \text{ cm}^{-3}$ from all the SIMS profiles are shown in figure 6.3. (see section 3.5.2 for details on the determination of junction abruptness and junction depth) The profiles are at least 10 nm deeper for the SOI substrate, which suggests a deeper melt depth for the same laser fluence used. On the other hand, improved junction abruptness with increasing laser fluence in the SOI substrates further confirms a greater melting duration, in which the dopant is allowed to redistribute more uniformly. (Further discussion on the melting duration in the SOI substrate will presented in the following section.) This indicates that the laser process has a strong dependence on the thermal conductivity of the substrate used [140].

Figure 6.3. Junction depth, determined at $1 \times 10^{18} \text{ cm}^{-3}$, and junction abruptness extracted from the SIMS profiles in figures 6.2 (a) and (b).

Figure 6.4. Cross-sectional TEM micrograph of 30-keV Si^+ preamorphized (a) silicon and (b) SOI substrates subjected to single-pulse LA at 0.6 J/cm^2 . The laser regrown layer in the SOI substrate is virtually defect free.

The cross-sectional TEM micrograph of the silicon substrate after laser irradiation at 0.6 J/cm^2 is shown in figure 6.4 (a). The melt front has apparently stopped near the amorphous/crystalline (a/c) interface. Since the laser fluence is unable to fully melt the PAI layer, the solidification process is mediated by explosive crystallization [110]. It is observed that at 0.6 J/cm^2 the $\sim 77 \text{ nm}$ thick amorphous silicon layer, shown in figure 6.1, resolidifies as polycrystalline silicon with a high density of stacking fault and microtwins, causing severe roughness to the surface and a/c interface as shown in figure 6.5. The epitaxial islands at the original a/c interface are believed to be nucleated from the roughened interface and defects induced by PAI.

Figure 6.5. Surface roughness of preamorphized Si and SOI samples subjected to single-pulse LA 0.2, 0.4 and 0.6 J/cm².

The XTEM image in figure 6.4 (b) for the SOI substrate shows that the SOI layer is a perfect crystal after the laser irradiation at 0.6 J/cm², and no defect can be found even at the original a/c region. This indicates that the thermal insulating effect of the BOX layer causes the melt front to exceed the PAI layer, unlike in the case of normal silicon substrate. The laser-regrown layer from this overmelting does not contain any defects which were observed in the absence of overmelting [55],[125]. Figure 6.5 shows the surface roughness of the preamorphized silicon and SOI samples subjected to single-pulse LA 0.2, 0.4 and 0.6 J/cm². The silicon substrate shows a consistent rougher surface than that of the SOI substrate. This is due the polycrystalline structure formed in the silicon substrate arising from the incomplete melting of the PAI layer.

Figure 6.6. SIMS profiles of boron in (a) silicon and (b) SOI substrates subjected to single-pulse LA at fluence of 0.45, 0.5, 0.6 and 0.63 J/cm².

6.2.2 LA of crystalline SOI and Si substrates

Figures 6.6 (a) and (b) compare the boron profiles in the crystalline silicon and the SOI substrates respectively subjected to single-pulse LA at laser fluences of 0.45, 0.5, 0.6

and 0.63 J/cm^2 . In the case of the crystalline silicon substrate, 0.6 and 0.63 J/cm^2 induce shallow substrate melting while no melting is induced at 0.45 and 0.5 J/cm^2 . The overlap of the laser annealed boron profiles at 0.45 and 0.5 J/cm^2 , shown in figure 6.6 (a), is an indication of an inadequate laser fluence to melt the crystalline silicon at the applied fluences. On the other hand, all fluences induce substantial melting in the SOI substrate. This indicates a significantly lower threshold of melting in the SOI substrate. From figure 6.7, the laser annealed boron profiles in the SOI substrate are at least 20 nm deeper than that in the silicon substrate. The difference in the junction depth is significantly larger than that in the preamorphized case, shown in figure 6.3. This suggests a stronger dependence on the thermal conductivity of the substrate than that when preamorphization technique is used.

Figure 6.7. Junction depth, determined at $1 \times 10^{18} \text{ cm}^{-3}$, and junction abruptness extracted from the SIMS profiles in figures 6.6 (a) and (b).

Figure 6.8. Comparison of boron profiles in Si and SOI substrates after single- and multiple-pulse LA at (a) 0.6 J/cm² and (b) 0.5 J/cm².

Figure 6.8 (a) compares the boron profiles in the silicon and the SOI substrates after single- or multiple-pulse LA at 0.6 J/cm². It can be seen that the boron profiles in the SOI substrate are significantly deeper than that in the silicon substrate for the same laser

fluences used. The junction depths formed in the SOI substrate by the 1-, 5- and 10-pulse irradiations are 113 nm, 140 nm and 150 nm respectively; while the corresponding junction depths in the silicon substrate are 39 nm, 39 nm and 41 nm respectively. A junction abruptness of ~ 7 nm/decade is found in both SOI and silicon substrates. Since the junction depth and abruptness are defined by the extent and the duration of the laser induced melting, a deeper melting would require a longer melt duration to uniformly distribute the dopant atoms. The results indicate a deeper melt depth and longer melt duration have occurred in the SOI substrate. It is also observed that with multiple-pulse LA, the junction depth in the SOI substrate increases with the subsequent laser pulses, a feature that is absent in the silicon substrate. It is postulated that the boron atoms did not diffuse to the maximum melt depth after the first pulse [24]. Upon the first laser irradiation, the melting of Si propagates into the solid at a velocity much higher than the boron diffusion velocity. Before the boron atoms reach the maximum melt depth, the molten Si resolidifies in the opposite direction of boron diffusion, causing the boron atoms to redistribute at a depth shallower than the maximum melt depth. As the number of laser pulses increases, the boron atoms distribute closer to the maximum melt depth.

As the laser fluence is reduced to 0.5 J/cm^2 , as shown in figure 6.8 (b), no significant melting is induced in the silicon substrate since there is no observable difference seen between the annealed profiles and the as-implanted profile upon laser irradiation. However, extensive melting in the SOI substrate is still observed at the same fluence of 0.5 J/cm^2 , indicating that the thermal confinement of the BOX layer lowers the threshold of melting. According to the law of heat conduction, the thermal energy transfer rate, H , through many layers of the materials in a steady state condition is given as [141]

$$H = \frac{A(T_2 - T_1)}{\sum_i \left(\frac{L_i}{k_i} \right)} \tag{Equation 6.1}$$

where k is the thermal conductivity, L is the thickness of the material and A is the cross sectional area. Based on equation 6.1, the calculated thermal energy transfer rate from the surface of the SOI substrate to a depth beyond the BOX layer will decrease over 97% in the thermal energy transfer rate as compared to that of a bulk silicon with the same substrate thickness. This signifies that a substantial thermal energy has been confined in the top silicon layer of the SOI substrate, leading to enhanced melting in the SOI substrate.

Figure 6.9. Sheet resistance of SOI and Si samples subjected to LA at 0.5 J/cm² and 0.6 J/cm² as a function of number of laser pulses.

From figure 6.9, it is noticed that the SOI samples have a lower sheet resistance than that of the silicon samples after the same annealing conditions. This is attributed to a higher level of activation and/or deeper junction depth formed in the SOI substrate. The sheet resistance values of the SOI samples remain relatively constant with increasing number of laser pulses, suggesting that the dopants are almost fully activated. For the silicon samples annealed in both melt and non-melt regimes, an increase in the number of laser pulses reduces the sheet resistance further without significantly increasing the junction depth. This is due to the fact that by increasing the number of pulses, a higher degree of activation can be achieved [26],[27],[119].

6.2.3 One-dimensional heat flow calculation in SOI substrate

Figure 6.10 shows the simulated maximum melt depths in the SOI and the silicon substrates as a function of laser fluence. Maximum melt depths extracted from the SIMS profiles in figures 6.8 (a) and (b) using the method described in section 3.5.2 are also shown for comparison. The simulation on the SOI substrate was conducted with a top SOI layer of 180 nm and a BOX layer of 180 nm. The simulated maximum melt depths of both SOI and silicon substrates show a good agreement with the extracted melt depths. The thermally less conductive BOX layer causes a significantly deeper melting in the SOI substrate as compared to that in the silicon substrate, also shown in figures 6.6 and 6.8. The maximum melt depths follow a linear relationship with the laser fluence upon laser irradiation at fluence greater than the threshold of melting. It is observed that the thermally less conductive BOX layer significantly reduces the threshold of melting of the SOI substrate, i.e. 0.25 J/cm^2 as compared to 0.52 J/cm^2 of the silicon substrate. Due to a

decrease in the thermal energy transfer rate, the laser irradiation quickly raises the temperature of the SOI substrate. This leads to increase in the laser-induced heating and melting of the SOI substrate.

Figure 6.10. Simulated maximum melt depths in the SOI and the silicon substrates as a function of laser fluence. Maximum melt depths extracted from the SIMS profiles in figures 6.8 (a) and (b) are also shown for comparison. The simulation on the SOI substrate was conducted with a top SOI layer of 180 nm and a BOX layer of 180 nm. (see also figure 4.5 for simulation on the silicon substrate)

Figure 6.11 indicates an increase in the melting rate (i.e. the slope of the leading edge of the melt front) in the SOI substrate, namely 11.23 m/s as compared to 5.32 m/s in the silicon substrate at 0.64 J/cm². The BOX layer also causes a decrease in the solidification rate (i.e. the slope of the trailing edge of the melt front) in the SOI substrate, namely 3.77 m/s as compared to 8.07 m/s in the silicon substrate at 0.64 J/cm².

The increase in the melting rate and the decrease in the solidification rate leads to an abrupt junction formation in the SOI substrate, as shown in figures 6.8 (a) and (b).

Figure 6.11. Simulated melt front upon laser irradiation of 23 ns FWHM starting at $t = 0$ on the SOI and the silicon substrates as a function of time at different laser fluences.

Figure 6.12 shows the simulated maximum melt depth in the SOI substrate as a function of varying top SOI and middle BOX thickness. The simulation was conducted with one varying thickness while another fixed at 180 nm. It is found that the maximum melt depth in the SOI substrate increases with increasing BOX thickness. This is attributed to a decrease in the thermal energy transfer rate due to a thicker BOX layer. An increase in the top SOI thickness, on the other hand, reduces the maximum melt depth in the SOI substrate. A thicker top SOI layer improves the thermal dissipation, reducing the thermal confinement effects induced by the thermally less conductive BOX layer.

Figure 6.12. Simulated maximum melt depth in the SOI substrate as a function of varying top SOI and middle BOX thickness. The simulation was conducted with one varying thickness while another fixed at 180 nm.

6.2.4 Non-melt LA in SOI substrate

The boron profiles in the SOI substrate after multiple-pulse, non-melt LA at 0.2 J/cm² are shown in figure 6.13. The fluence chosen is sufficiently lower than the calculated threshold of melting of the SOI substrate to prevent possible melting due to typical laser fluctuation. It can be seen that the boron profiles after the LA of 1-, 10- and 20-pulse irradiations overlap with the as-implanted profile. This suggests that no bulk melting has occurred with these annealing conditions, forming ultra-shallow junctions defined by the as-implanted profile. However, when the number of pulses is increased to 50 pulses, significant diffusion of boron is observed, and this suggests that melting has taken place. It is believed that the feature is caused by a surface roughness enhanced absorption of the laser irradiation [119].

Figure 6.13. Boron profiles in SOI substrate subjected to LA at 0.2 J/cm² with different number of laser pulses.

Figure 6.14. Surface roughness of Si and SOI samples subjected to multiple-pulse non-melt LA at 0.5 J/cm² and 0.2 J/cm², respectively.

AFM analysis shows an increase in the root-mean-square (RMS) surface roughness, i.e. 1.34 nm and 2.68 nm at the 10- and the 50-pulse conditions shown in figure 6.14. Top-view and three-dimensional AFM images of the SOI sample subjected to the 50-pulse laser irradiation at 0.2 J/cm^2 are shown in figures 6.15 (a) and (b), respectively. This suggests that the thermal confinement of the BOX layer induces a sufficiently high temperature, promoting surface roughening at a given number of pulses (i.e. more than 20 pulses in this particular case). Since LA was conducted in normal ambience, the possibility of oxygen reacting with the Si surface cannot be dismissed [120]. The surface becomes rougher with an increase in the number of pulses, which in turn increases the absorption of the laser irradiation [119]. Furthermore, the surface melting at the 50-pulse LA can also cause surface roughening after rapid rearrangement of Si atoms at the surface upon solidification [120]. The increase in the surface roughness after LA for samples subjected to 0.2 J/cm^2 at 50 pulses, as shown in figure 6.15, may degrade the electrical characteristics of MOSFET. Leakage current can be significant, resulting from an uneven distribution of junction depth and non-uniform silicide formation [2]. It is important to note that the surface roughness after the 50-pulse laser irradiation is sufficiently high to degrade the SIMS depth resolution. The true junction profile should therefore be shallower than that reflected in the SIMS profile shown in figure 6.13.

Figure 6.16 depicts the sheet resistance of the SOI samples after single- and multiple-pulse LA at 0.2 J/cm^2 . The sheet resistance decreases as the number of pulses increases, implying a higher activation is resulted. The sheet resistance decreases from $588 \text{ } \Omega/\square$ to $429 \text{ } \Omega/\square$ when the number of laser pulses increases from 20 to 50 pulses,

which is accompanied by an increment in the junction depth. This suggests that the maximum activation has reached after the 20-pulse irradiation for the non-melt LA.

Figure 6.15. (a) Top-view and (b) three-dimensional AFM images of SOI sample subjected to 50-pulse laser irradiation at 0.2 J/cm^2 .

Figure 6.16. Sheet resistance of SOI samples annealed at 0.2 J/cm² as a function of number of laser pulses. The sheet resistance after LA of 0.5 J/cm² and 0.6 J/cm² are included for comparison.

It should be noted that the thermal confinement effect induced by the thermally less conductive BOX layer during LA also affects the junction formation in the lateral direction, especially adjacent to/underneath the gate edge. This poses a great challenge in controlling the lateral dopant diffusion in the SOI substrate during shallow-melt LA. Non-melt LA, on the other hand, produces a diffusionless active dopant profile (in both vertical and lateral directions) defined by the as-implanted profile. Non-melt LA is an attractive process in the junction formation in SOI substrate for its improved dopant activation and controlled dopant profile (in both vertical and lateral directions) defined by as-implanted profile.

6.3 Thermal Confinement in Strained-Si/SiGe Substrate during LA

Ultralow energy B⁺ was implanted at 0.2 keV with a dose of $2.5 \times 10^{14} \text{ cm}^{-2}$ and $5 \times 10^{14} \text{ cm}^{-2}$ on n-type Si (100) and strained-Si/SiGe substrates. The strained-Si/SiGe substrate comprises a linearly graded SiGe layer 2 μm thick grown on p-type Si wafers. On top of the graded SiGe, 1 μm of uniform SiGe layer at 25% Ge content and 20 nm of strained Si layer were grown. Laser annealing was carried out in normal ambience using a 248 nm KrF excimer laser with a pulse duration of 23 ns and a repetition rate of 1 Hz.

Figures 6.17 (a) and (b) show the boron and the germanium profiles in the bulk Si and the strained-Si/SiGe substrates, respectively, subjected to single-pulse LA at 0.6 J/cm² and multiple-pulse LA at 0.4 J/cm². In the case of the bulk Si substrate, the LA at 0.6 J/cm² induces a shallow melting of 28 nm while 0.4 J/cm² induces no melting to the substrate. An abrupt boron profile at 0.6 J/cm² is contained in the melted region. The overlapping of the annealed profiles at 0.4 J/cm² with the implanted profile is an indication that the laser fluence is inadequate to melt the crystalline Si. Using a one-dimensional heat flow calculation [92],[96] and SIMS analysis, a threshold of melting of 0.52 J/cm² is derived for the bulk Si substrate. On the other hand, all fluences, including the low fluence LA of 0.4 J/cm², induce substantial melting in the strained-Si/SiGe substrate, as shown in figure 6.17 (b). The profiles are at least 100 nm deeper in the strained-Si/SiGe substrate than that in the bulk Si substrate. This indicates a substantial lowering of the threshold of melting in the strained-Si/SiGe due to a thermal confinement effect of the SiGe layer. A threshold of melting of 0.15 J/cm² is found for the strained-Si/SiGe substrate. A graded boron profile at a single-pulse LA at 0.4 J/cm² is resulted due to a deep melted layer that boron could not redistribute fast enough to form a box-like

profile. The germanium step at the original strained-Si/relaxed-SiGe interface shown in figure 6.17 (b) suggests a possible melting of the relaxed SiGe layer prior to the strained-Si layer due to a lower melting temperature of SiGe than that of Si [142]. Box-like profiles of boron and uniform germanium profiles are generated at 5- and 10-pulse LA at 0.4 J/cm^2 due to a longer melting duration that allows a uniform distribution of boron and germanium atoms.

As the thermal conductivity of the SiGe layer is very much less than Si [142], the SiGe layer in the strained-Si/SiGe substrate has an insulating effect that prevents heat dissipation. Based on equation 6.1, the calculated thermal energy transfer rate from the surface of the strained-Si to a depth beyond the graded-SiGe layer will decrease over 94% in the thermal energy transfer rate as compared to that of a bulk Si with the same substrate thickness. When applying LA with pulse duration in nanoseconds to substrates such as the strained-Si/SiGe, one would expect a significant difference in the heating and cooling rates as compared to a silicon substrate.

Figure 6.18 (a) shows the SIMS profiles of boron and germanium in the strained-Si/SiGe substrate subjected to a shallow-melt LA as low as 0.2 J/cm^2 . The melting mechanism is similar to the case of 0.4 J/cm^2 , as shown in figure 6.17 (b). This causes a great challenge in creating a shallow melting of the strained-Si without melting the SiGe layer that has a lower melting temperature than that of Si. Non-melt LA shows to be the best alternative for the activation of junction [26],[27].

Figure 6.17. SIMS profiles of boron in (a) silicon and (b) strained-Si/SiGe substrates subjected to single and multiple-pulse LA at 0.4 J/cm² and 0.6 J/cm². Germanium profiles are also included in (b).

Figure 6.18. SIMS profiles of boron and germanium in strained-Si/SiGe substrate subjected to (a) shallow-melt LA at 0.2 J/cm² and (b) non-melt LA at 0.1 J/cm².

Figure 6.18 (b) shows the boron and the germanium profiles subjected to a non-melt LA at 0.1 J/cm². The complete overlapping of the boron and the germanium profiles with the as-implanted/as-grown profiles confirms no melting occurs during the LA. No

germanium out-diffusion, which would affect the performance of the strained-Si layer, is observed for the 23-ns laser pulse. A good activation in the non-melt LA is indicated by the low sheet resistance shown in figure 6.19. A further decrease in the sheet resistance can be achieved with increasing laser pulses. A “zig-zag” in the sheet resistance at higher fluences represents a non-uniformity of germanium content upon melting induced by the initial laser pulses.

Figure 6.19. Sheet resistance of strained-Si/SiGe samples annealed with LA conditions specified in figures 6.17 and 6.18. Each box plots represents four measured data points.

The Raman spectra of the strained-Si/SiGe substrate, shown in figure 6.20, indicate an absence of the Si-Si peak in the strained-Si layer at 0.2 J/cm². This indicates that the melting at 0.2 J/cm² (see also figure 6.18 (a)) has consumed the strained-Si and relaxed-SiGe layers, leading to a distribution of germanium into the original strained-Si layer. On the other hand, no degradation in the strain in the strained-Si layer was observed after the non-melt LA at 0.1 J/cm² as shown by the similar Raman spectra to that of the as-grown

sample. Therefore, non-melt LA proves to be a promising alternative for annealing advanced semiconductor substrate with a thermally less conductive substrate.

Figure 6.20. Raman spectra of strained-Si/SiGe samples annealed at 0.2 J/cm² (shallow-melt LA) and 0.1 J/cm² (non-melt LA).

6.4 Summary

The results indicate that LA has a strong dependence on the thermal conductivity of the substrate used. This causes a great challenge in controlling the melt depth in the SOI and the strained-Si/SiGe substrates during shallow-melt LA. We have shown that non-melt LA produces dopant profiles of negligible diffusion and improved activation in both SOI and strained-Si/SiGe substrates. Furthermore, negligible degradation in the strain of

the strained-Si layer of the strained-Si/SiGe substrate is induced with increasing the number laser pulses at non-melt LA. Non-melt LA is an attractive process in the dopant activation of thermally less conductive semiconductor substrates.

CHAPTER 7

FORMATION OF SELF-DEFINED NICKEL SILICIDED

ULTRA-SHALLOW P^+/N JUNCTION USING LASER

ANNEALING

7.1 Introduction

LA has received a renewed interest in the formation of advanced metal silicides in recent years. LA provides an ultra rapid heating and quenching that is capable forming epitaxial metal silicides through LPE solidification. A smooth and abrupt silicide/Si interface, which leads to excellent junction leakage characteristics, can be achieved through the melting and resolidification induced by LA. In this chapter, we present a simultaneous formation of Ni-silicided p^+/n ultra-shallow junction using a single-pulse LA.

7.2 Self-Defined p^+/n Ultra-Shallow Junctions

B^+ was implanted at 0.5 keV with a dose of $1 \times 10^{15} \text{ cm}^{-2}$ on an n-type (100) bulk silicon substrate. The samples were cleaned using a standard Radio Corporation of America (RCA) solution and diluted HF prior to the deposition of a 15-nm Ni layer. LA was then performed to simultaneously activate the junction and form Ni-silicide. Isolated by shallow trench, p^+/n diodes were also fabricated using the same implantation, Ni layer and LA conditions mentioned above. pMOSFET with a thermally grown 6.5 nm gate oxide was also fabricated for the electrical study. After gate stack patterning, the source/drain region was formed using a B^+ implantation at 2.0 keV with a dose of 3×10^{15}

cm^{-2} . The 2.0-keV B^+ implantation creates a deeper vertical junction to prevent excessive junction leakage current at the uncompensated gate/spacer edge. Unreacted Ni layer was removed using a standard solution at 60 °C.

Figure 7.1. SIMS profiles of (a) Ni and (b) B subjected to a single-pulse LA at 0.3 J/cm², 0.4 J/cm² and 0.5 J/cm². The dotted lines in (b) denote the Ni-silicide/Si interfaces that are defined at a Ni concentration (counts) of one order lower from the edge of flat profile.

Figures 7.1 (a) and (b) show the SIMS profiles of Ni and B, respectively, subjected to a single-pulse LA at 0.3 J/cm², 0.4 J/cm² and 0.5 J/cm². It is found that the Ni and B diffusion into the Si substrate increases with the increasing laser fluences. The Ni diffusion into the Si substrate shows a linear increase with the laser fluence (see figure

7.2), due to a higher laser-induced temperature with elevating fluences. As shown in figure 7.1 (a), a well-defined Ni-silicide/Si interface due to the rapid heating and quenching of the laser irradiation is seen at all fluences. (The compositional analysis of the Ni-silicide layer will be discussed in the following section.) As depicted in figures 7.1 (b) and 7.2, the B diffusion shows a similar linear relationship at fluences greater than 0.3 J/cm². No significant diffusion in the B profile tail (below a concentration of 5×10¹⁷ cm⁻³) is observed for the LA at 0.3 J/cm². The redistribution of B atoms in the Ni-silicide layer and adjacent to the Ni-silicide/Si interface is observed for LA at 0.3 J/cm². This B redistribution is caused by the Ni diffusion during LA. (The B and Ni diffusion during LA is addressed in the following section.)

Figure 7.2. Thickness of the Ni-silicide layer and junction depth (defined at concentration of 1×10¹⁸ cm⁻³) underneath the Ni-silicide layer upon LA extracted from the results shown in figure 7.1. Results for 0.2 J/cm² annealed sample are also included for comparison.

Figure 7.3 shows the simulated temperatures at different depths in a 15-nm Ni/Si system as a function of laser fluence. The simulations were conducted using a one-dimensional heat flow calculation [92],[96] on the 15-nm Ni/Si system. Simulation was also performed on a bare Si substrate without Ni layer. It is found that the enhanced optical coupling of the Ni layer increases the surface temperature (i.e. at $x = 0$ nm in the one-dimensional axis) by as much as $400\text{ }^{\circ}\text{C}$ at 0.5 J/cm^2 as compared to that of the bare Si substrate. The resulted temperature in the Ni/Si system (i.e. $x = 10$ nm in the Ni layer and $x = 40$ nm in the underlying Si substrate) at a fluence of 0.4 J/cm^2 and above is sufficiently high to melt both Ni and Si layers. Since the simulation does not consider the dynamic Ni-Si intermixing that changes the optical and thermal properties of the Ni/Si system during laser irradiation, a higher temperature is expected in the empirical system upon the laser irradiation. Ni and B atoms experience a fast diffusion in liquid Si for laser fluences greater than 0.4 J/cm^2 , leading to a significantly thicker Ni-silicide layer and a deeper junction as shown in figures 7.1 (a) and (b). On the other hand, the negligibly small amount of diffusion of B atoms in the profile tail at 0.3 J/cm^2 , as shown in figure 7.1 (b), is caused by a slow solid-phase B diffusion in the Si substrate upon LA. Since the laser irradiation only lasts for tens of nanoseconds, B atoms would only experience a negligible diffusion in the profile tail as seen for laser fluences lower than 0.3 J/cm^2 . As shown in figure 7.2, an increase in the laser fluences from 0.2 J/cm^2 to 0.3 J/cm^2 leads to a decrease in the junction depth (i.e. defined at the concentration of $1 \times 10^{18}\text{ cm}^{-3}$ underneath the Ni-silicide layer). This is attributed to the diffusionless B profile tail (less than the concentration of $1 \times 10^{18}\text{ cm}^{-3}$) and an increase in the Ni-silicide layer thickness upon the increase in the laser fluences from 0.2 J/cm^2 to 0.3 J/cm^2 . As the B profile tail

does not diffuse significantly at 0.2 J/cm² and 0.3 J/cm² due to solid-phase B diffusion, the Ni diffusion into the Si substrate consumes part of the B profile, leading to a shallow junction underneath the Ni-silicide layer. A junction as shallow as 16 nm was formed at 0.3 J/cm² through the described mechanism. A pile-up of B atoms is also seen adjacent to the Ni-silicide/Si interface in the Si substrate at all fluences, as shown in figure 7.1 (b). This is due to a higher diffusivity of B in Ni-Si layer and a higher segregation efficiency of B at the Ni-silicide/Si interface [143],[144]. This suggests a “push-forward” effect of the Ni diffusion and/or the Ni-silicide formation on the B atoms, ensuring that the junctions are always deeper than the Ni-silicide layer. The “push-forward” effect also produces a good junction abruptness at low fluence, such as 5 nm/decade at 0.3 J/cm².

Figure 7.3. Simulated temperature at different depths in a 15-nm Ni/Si system as a function of laser fluence. The depth assumes a direction into the Si substrate, e.g. $x = 0$ nm represents the surface of Ni layer. The simulated surface temperature of a bare Si substrate without Ni layer (i.e. $x = 0$ nm) is also included for comparison.

Figures 7.4 (a) and (b) show the SIMS profiles of Ni and B, respectively, subjected to a single- and a 10-pulse LA at 0.3 J/cm^2 and 0.5 J/cm^2 . The Ni profile deepens slightly upon the 10-pulse LA at 0.5 J/cm^2 . Upon the initial laser irradiation, the Ni-silicide layer of lower melting temperature (as compared with pure Ni layer) [145] is formed. This Ni-silicide layer limits the further melting induced by the subsequent laser irradiations. The subsequent laser irradiations melt the Ni-silicide/Si system up to the Ni-silicide/Si interface due to the melting temperature difference between the Ni-silicide and the Si layers. Nevertheless, the high temperature induced at 0.5 J/cm^2 causes a melting slightly beyond the Ni-silicide/Si interface, leading to a slight increase in the Ni-silicide layer thickness. Despite that the subsequent laser irradiations melts only the Ni-silicide layer that limits the Ni diffusion, a substantially high temperature is induced at 0.5 J/cm^2 . This results in a significant B diffusion at the 10-pulse LA at 0.5 J/cm^2 , as shown in figure 7.4 (b). The 10-pulse LA at 0.3 J/cm^2 , on the other hand, results in a similar Ni-silicide layer thickness and junction depth underneath the Ni-silicide layer to that of the single-pulse LA at 0.3 J/cm^2 . The lower temperature induced by the 10-pulse LA at 0.3 J/cm^2 is insufficient to cause any over-melting of the Ni-silicide layer and significant B diffusion. The B pile-up adjacent to the Ni-silicide/Si interface increases with increasing laser pulse at both 0.3 J/cm^2 and 0.5 J/cm^2 , as shown in figure 7.4 (b). The results show an increase in the pile-up peak concentration by 1.4 and 2.3 times for the 10-pulse LA at 0.5 and 0.3 J/cm^2 , respectively. The high temperature induced by the 10-pulse LA at 0.5 J/cm^2 leads to the broadening of the B pile-up. The 10-pulse LA at 0.3 J/cm^2 , on the other hand, causes little B diffusion in the profile tail under a lower induced temperature and leads to a higher concentration of B pile-up adjacent to the Ni-silicide/Si interface. The high

concentration of the B pile-up near the Ni-silicide/Si interface would serve as low resistive contact between the Ni-silicide and the junction [146].

Figure 7.4. SIMS profiles of (a) nickel and (b) boron subjected to a single- and 10-pulse LA at 0.3 J/cm² and 0.5 J/cm². The dotted lines in (b) denote the Ni-silicide/Si interfaces. 1P and 10P denote single- and 10-pulse LA, respectively.

7.3 Electrical Characteristics of Ni Silicided Ultra-Shallow Junctions

Figure 7.5 shows the diode characteristics of Ni-silicided p^+/n junctions formed by multiple-pulse LA at 0.5 J/cm², 0.4 J/cm², 0.3 J/cm² and 0.2 J/cm². It is found that the enhanced optical coupling of the as-deposited Ni layer has generated a substantially high temperature, leading to a high degree of activation of B atoms. As shown in figures 7.5

(a) and 7.6, the junction leakage current remains low for the multiple-pulse LA at 0.5 J/cm². This is due to a deeper junction formed upon multiple-pulse (see figure 7.4 (b)). The multiple-pulse LA at 0.4 J/cm² produces low leakage junction up to a laser irradiations of 10 pulses. As the junction formed at 0.4 J/cm² lies considerably nearer to the Ni-silicide/Si interface, a large number of the laser pulses would cause a certain amount of Ni atoms to diffuse into the depletion region of the junction, leading to a higher junction leakage current (as seen in the 100-pulse laser annealed diodes).

Figure 7.5. Junction leakage characteristics of p^+/n junctions subjected to multiple-pulse LA at (a) 0.5 J/cm² and 0.4 J/cm² and at (b) 0.3 J/cm² and 0.2 J/cm². 1P, 10P and 100P denote single-, 10- and 100-pulse LA, respectively.

The junction leakage current becomes more dominant at low fluence 100-pulse LA of 0.3 J/cm² and 0.2 J/cm², as shown in figure 7.5 (b), where the junctions are nearer to the Ni-silicide layer. A 10-pulse LA at 0.3 J/cm² and 0.2 J/cm² (also shown in figure 7.6)

reduces the junction leakage current through an annihilation of point defects generated during implantation and left unannealed at the initial laser irradiation. It should be noted that diodes formed without the B implantation (i.e. with only a 15-nm Ni layer) shows no diode characteristics upon LA.

Figure 7.6. Junction leakage current of p^+/n junctions (in figure 7.5) at -1.6 V as a function of number of laser pulses irradiated.

Figures 7.7 (a) and (b) show the I_d - V_{ds} and I_d - V_{gs} characteristics, respectively, of a $0.5\text{ }\mu\text{m}$ pMOSFET subjected to a single-pulse LA at 0.4 J/cm^2 and 0.3 J/cm^2 . As shown in figure 7.7 (a), the LA at 0.4 J/cm^2 produces a higher drive current as compared to that at 0.3 J/cm^2 due to an enhanced B activation at higher fluence. A higher leakage current of about one order is observed for the LA at 0.3 J/cm^2 , as shown in figure 7.7 (b). This is because that the p^+/n junction is too near to the Ni-silicide layer, as shown in figures 7.1 (a) and (b). The inset in figure 7.7 (b) also shows a lower transconductance at the LA of

0.3 J/cm² as compared to that of 0.4 J/cm². The lower drive current produced by the LA of 0.3 J/cm² is also attributed to the formation of a high resistive phase of Ni-silicide layer in the source/drain and gate regions, which will be discussed in the following section. The high resistive phase of Ni-silicide leads to a higher parasitic resistance.

Figure 7.7. (a) Drain current (I_d) versus drain bias (V_{ds}) and (b) drain current (I_d) versus gate bias (V_{gs}) characteristics of a 0.5 μm pMOSFET subjected to a single-pulse LA at 0.4 J/cm² and 0.3 J/cm². The inset in (b) shows the transconductance of the 0.5 μm pMOSFET.

7.4 Laser Induced Ni Silicide Formation

Figure 7.8 shows the XRD spectra of the 15-nm Ni/Si sample subjected to single- and 10-pulse LA at 0.3 J/cm² and 0.4 J/cm². It is found that the single- and 10-pulse LA at 0.3 J/cm² did not induce any silicidation. On the other hand, NiSi₂ formation is observed for the single- and 10-pulse LA at 0.4 J/cm². The NiSi₂ peak intensity is higher

for the 10-pulse laser annealed sample, signifying an enhanced silicidation at higher laser pulses. Figure 7.9 shows the XTEM micrograph of the 15-nm Ni/Si sample subjected to the single-pulse LA at 0.4 J/cm^2 . A distinctive 45-nm layered structure consisting of an amorphous Ni-Si layer and a poly-crystalline NiSi_2 [147] is observed. The thickness of the layered structure agrees well with the SIMS profile thickness of 44 nm, as shown in figure 7.1 (a). The formation of the layered structure is due to a supercooling experienced by the melted region [147]. In the initial molten state, the Ni, Si and B atoms intermix in a liquid phase up to a maximum melt depth of 44 nm. It is believed that the molten layer at the liquid/Si interface is Si-rich and becomes Ni-rich as it goes towards the surface. Upon solidification, the single crystalline Si substrate at the liquid/solid interface facilitates the nucleation of NiSi_2 which has a very similar crystal structure and lattice spacing with Si. As the melt front moves towards the surface, the Ni-rich Ni-Si molten layer experiences a supercooling and rapidly quenched which is due to the fact that Ni-rich Ni-Si compound has a higher solidification temperature [145]. The supercooled liquid falls rapidly below the solidification temperature for amorphous Ni-silicide, leading to the formation of an amorphous Ni-rich Ni-Si layer at the surface. The high resistive surface amorphous Ni-Si layer causes a high sheet resistance measured, as shown in figure 7.10. Although the XRD spectra (see figure 7.8) show no Ni-silicide formation at the single- and 10-pulse LA at 0.3 J/cm^2 , the SIMS profile in figures 7.1 (a) and 7.3 (a) show a diffusion of Ni atoms into Si substrate. The intermixing of Ni and Si atoms does not form Ni-silicide and remains as a high resistive Ni-Si intermixed layer, as seen in figure 7.10. The sheet resistance, as shown in figure 7.10, decreases with

increasing laser fluence due to a thicker Ni-silicide layer and the formation of NiSi_2 at high fluences, such as at 0.4 J/cm^2 and 0.5 J/cm^2 .

Figure 7.8. XRD spectra of a 15-nm Ni/Si sample subjected to single- and 10-pulse LA at 0.3 J/cm^2 and 0.4 J/cm^2 .

Figure 7.9. XTEM micrograph showing a formation of a layered structure, consisting of an amorphous Ni-rich Ni-Si overlayer and a poly-crystalline NiSi_2 layer, from a 15-nm Ni/Si sample subjected to a single-pulse LA at 0.4 J/cm^2 .

Figure 7.10. Sheet resistance of single- and 10-pulse laser annealed Ni/Si samples as a function of laser fluence.

7.5 Ni-Silicided p^+/n Ultra-Shallow Junctions upon Post-LA Low Temperature RTA

The previous section shows the formation of NiSi₂ at high laser fluence, i.e. greater than 0.4 J/cm². Despite that NiSi₂ has a low resistivity of 50 μΩcm, the final sheet resistance does not meet the stringent ITRS silicide requirement for shorter devices [4]. NiSi phase, of which is typically formed at a lower RTA temperature than that of NiSi₂ phase, has a lower resistivity of 15 μΩcm. Low resistivity NiSi (which is also free from sheet resistance degradation at high temperature) is capable of meeting the stringent ITRS silicide requirement for sub-45nm MOSFET. This section studies the ultra-shallow junction and NiSi formation caused by a post-LA RTA at 400 °C on the 0.3 J/cm² laser annealed Ni-capped sample.

Figures 7.11 (a) and (b) show the SIMS profiles of Ni and B, respectively, subjected to a single-pulse LA at 0.3 J/cm^2 with or without a post-LA RTA at 400°C for 30 s. It is observed that the LA at 0.3 J/cm^2 causes a diffusion of Ni atoms into the Si substrate within a duration of tens of nanoseconds, creating a well-defined Ni-silicide/Si interface. The subsequent RTA at 400°C for 30 s, on the other hand, causes negligible diffusion to the Ni profile. This is due to a pre-intermixing of Ni and Si atoms upon the initial LA and a formation of NiSi toward the surface upon subsequent RTA. (Further discussion on NiSi formation will be covered in the following section.) Because of a low B diffusivity in solid Si, the B profile tail in figure 7.11 (b) also experiences a negligible diffusion upon LA despite a high temperature of about 960°C (simulated by the one-dimensional heat flow calculation) induced by the optical coupling of the Ni layer, as shown in figure 7.3.

The subsequent RTA at 400°C is also insufficient to causes any significant B diffusion in the Si substrate. The subsequent RTA, however, causes a high concentration of B to be segregated at the Ni-silicide/Si interface and at Ni-silicide surface. The high diffusivity of B in the Ni-silicide layer and high segregation efficiency at the interface cause the initial B pile-up adjacent to the Ni-silicide/Si interface [143],[144]. Upon the post-LA RTA, B atoms move rapidly in the Ni-silicide layer and segregate at the Ni-silicide/Si interface and at Ni-silicide surface. It is believed that the movement of B atoms is further enhanced by the formation of NiSi. This leads to an increase of B pile-up adjacent to the Ni-silicide/Si interface by 1.4 times by the subsequent RTA. The high concentration of B pile-up adjacent to the Ni-silicide/Si interface would serve as a low resistive contact between the Ni-silicide and the junction [146].

Figure 7.11. SIMS profiles of (a) Ni and (b) B subjected to a single-pulse LA at 0.3 J/cm^2 with or without a post-LA RTA at 400°C for 30 s. The dotted lines in (b) denote the Ni-silicide/Si interfaces.

Figure 7.12 shows the diode characteristics of Ni-silicided p^+/n junctions formed by the single-pulse LA at 0.3 J/cm^2 with or without the post-LA RTA. Due to the enhanced optical coupling induced by the Ni layer, high temperature created by LA leads to the activation of the B atoms (refer to section 7.2). It should be noted that the diode formed with only Ni layer (i.e. without B implantation) shows no diode characteristics, indicating that the diode characteristics is created by the activation of B atoms underneath the formed Ni-silicide layer. It is found that a higher junction leakage is resulted upon the

post-LA RTA. As the junction formed at 0.3 J/cm^2 lies considerably near to the Ni-silicide/Si interface, the post-LA RTA would cause a certain amount of Ni atoms to diffuse into the depletion region of the junction, leading to a higher junction leakage current.

Figure 7.12. Diode characteristics of Ni-silicided p^+/n junctions formed by single-pulse LA at 0.3 J/cm^2 with or without the post-LA RTA at 400°C for 30 s.

Figure 7.13 (a) shows the XTEM micrograph of the 15-nm Ni/Si sample subjected to the single-pulse LA at 0.3 J/cm^2 . It is observed that the LA at 0.3 J/cm^2 creates an intermixing of Ni and Si atoms in a continuous amorphous Ni-Si layer with a smooth film/Si interface (also see in figure 7.11 (a)). The XRD spectra in figure 7.14 indicate no Ni-silicide formation at the single-pulse LA at 0.3 J/cm^2 despite diffusion of Ni atoms. Upon laser irradiation on the Ni/Si sample, high temperature of about 960°C is induced,

leading to a fast diffusion of Ni atoms into the Si substrate up to a depth of about 30 nm as shown in figure 7.13 (a). Rapid quenching or cooling of LA towards the surface then generates an amorphous Ni-Si layer. The high resistive amorphous Ni-Si layer results in a high sheet resistance of $73 \Omega/\square$, as shown in figures 7.10 and 7.15.

Figure 7.13. XTEM micrographs of the 15-nm Ni/Si sample subjected to (a) single-pulse LA at 0.3 J/cm^2 and (b) single-pulse LA at 0.3 J/cm^2 followed by a RTA at 400°C for 30 s. The inset in (b) shows a layered structure of NiSi at the film/Si interface and amorphous Ni-Si overlayer.

The post-LA RTA, on the other hand, causes a formation of NiSi phase as shown in figures 7.13 (b) and 7.14. The inset in figure 7.13 (b) shows a layered structure of NiSi at the film/Si interface and amorphous Ni-Si overlayer. The unreacted amorphous Ni-Si surface layer is believed to be the result of incomplete NiSi formation from the initial amorphous Ni-Si layer induced by LA. This indicates that a NiSi formation was

originated at the film/Si interface and moved towards to the surface [147]. The formation of the NiSi phase results in a low sheet resistance of $9 \, \Omega/\square$, as shown in figure 7.15. It should be noted that high fluence LA greater than $0.4 \, \text{J}/\text{cm}^2$ followed by the RTA at $400 \, ^\circ\text{C}$ for 30 s leads to a further formation of NiSi_2 from the NiSi_2 template created by the initial LA at the film/Si interface (see section 7.4). This leads to a higher sheet resistance as compared to that of LA at $0.3 \, \text{J}/\text{cm}^2$ followed by the RTA, as shown in figure 7.15.

Figure 7.14. XRD spectra of the 15-nm Ni/Si sample subjected to single-pulse LA at $0.3 \, \text{J}/\text{cm}^2$ with or without the post-LA RTA at $400 \, ^\circ\text{C}$ for 30 s.

Figure 7.15. Sheet resistance of single-pulse laser annealed Ni/Si sample with or without the RTA at 400 °C for 30 s as a function of laser fluence.

7.6 Summary

Enhanced optical coupling of the as-deposited Ni layer on Si substrate generates a substantially high temperature upon LA that is capable of simultaneously activating the ultra-shallow junction and forming a Ni-silicide layer. The extent of Ni and B diffusion is determined by the laser fluence. The self-defined ultra-shallow junction is always deeper than the Ni-silicide layer due to a “push-forward” effect induced by the Ni diffusion. Despite that the junctions lie near to the Ni-silicide layer, good diode and transistor characteristics are produced. A distinctive layered structure consisting of an amorphous Ni-rich Ni-Si overlayer and a poly-crystalline NiSi₂ layer at the film/Si interface is formed at high fluences due to a supercooling of the melted region. Low laser fluence of 0.3 J/cm², on the other hand, causes intermixing of Ni and Si atoms, but does not induce

significant Ni silicidation. A post-LA RTA at 400 °C, 30 s causes NiSi formation in the 0.3 J/cm² laser annealed sample. The post-LA RTA does not lead to significant diffusion to the laser annealed Ni profile and B profile tail.

CHAPTER 8

CONCLUSIONS AND RECOMMENDATIONS

8.1 Conclusions

LA has been shown to be a promising technique in the formation of ultra-shallow junction for nano-scale MOSFET technologies. Shallow-melt LA melts the surface region of the Si substrate, producing a highly activated abrupt junction through rapid redistribution of dopant atoms in liquid Si. A preamorphization implantation, such as Si^+ or Ge^+ , can be used to enhance the process window of shallow-melt LA. However, a melt-depth related dopant loss could compromise the advantages provided by shallow-melt LA. Non-melt LA, on the other hand, produces a moderately activated junction with negligible dopant diffusion or loss to the as-implanted dopant profile. Multiple-pulse LA has been shown to be effective in improving the dopant activation without compromising the junction depth and the junction abruptness. The maximum number of laser pulses that can be applied during LA has to be controlled to prevent surface degradation induced by multiple-pulse LA. It is also found that the combination of ultra-low energy implantation and multiple-pulse non-melt LA is suitable for nano-scale MOSFET technologies.

A Si^+ subamorphization implantation can be used to further enhance the dopant activation upon non-melt LA. The enhancement is caused by a vacancy-rich surface region generated by the Si^+ subamorphization implantation that promotes the incorporation of boron atoms into the substitutional sites. The additional Si^+ subamorphization implantation does not alter the threshold of melting and produces a comparable junction leakage current.

LA produces a high dopant activation that exceeds the equilibrium dopant solid solubility in the Si substrate. This high dopant activation is in a metastable state and could be deactivated upon subsequent thermal cycles. A substantial dopant deactivation takes place for post-LA RTA temperatures greater than 800 °C in Si^+ and Ge^+ preamorphized samples. This dopant deactivation arises from the presence of excess interstitials in the end-of-range region that could not be annealed out by melting up to the amorphous/crystalline interface. Melting beyond the amorphous layer reduces the amount of excess interstitials in the end-of-range region and suppresses the dopant deactivation. Reduced dopant deactivation and dopant diffusion upon post-LA RTA is observed with the Ge^+ PAI scheme.

Melting of Si^+ preamorphization implantation layer down to or just beyond the a/c interface by pulsed LA may not be sufficient to remove the crystalline defects generated during the implantation and the fast LPE regrowth, resulting in a high leakage current despite a highly activated p^+/n junction. A post-LA low temperature RTA of 600 °C has been shown to be critical for the further elimination of the crystalline defects in the laser annealed preamorphized devices. A process combination of LA and the post-LA low temperature RTA produces good electrical device characteristics.

LA has a strong dependence on the thermal properties of the advanced semiconductor substrates, such as silicon-on-insulator and strained-silicon/SiGe substrates. A thermal confinement effect of enhanced heating and melting is induced by the thermally less conductive layers in the silicon-on-insulator and the strained-silicon/SiGe substrates. This causes a great challenge in controlling the melt depth in silicon-on-insulator and strained-silicon/SiGe substrates during shallow-melt LA. The

results indicate that non-melt LA is a suitable technique in producing highly activated diffusionless ultra-shallow junction in the silicon-on-insulator and the strained-Si/SiGe substrates.

It should be noted that high fluence shallow-melt LA produces high dopant activation and improves junction leakage. However, high fluence shallow-melt LA also causes undesirable effects on MOSFET, such as physical gate degradation, deeper vertical and lateral junction depths. The control over the lateral dopant distribution (at the gate edge) induced by shallow-melt LA complicates due to a melting dependence on the optical and the thermal properties of the gate at the gate edge. Low fluence non-melt LA, on the other hand, do not induced significant melting on the gate and produces a diffusionless dopant profile (both vertical and lateral) defined by the as-implanted profile. This indicates that non-melt LA is a superior technique for the formation of ultra-shallow junction in MOSFET.

A capped Ni layer on the Si substrate generates a substantially higher temperature upon LA due to an enhanced optical coupling. The induced high temperature simultaneously activates the ultra-shallow junction and forms a Ni-silicide layer. The extent of Ni and B diffusion is determined by the laser fluence. The technique demonstrates good junction leakage and transistor characteristics. High fluence LA leads to the formation of a distinctive layered Ni-silicide structure consisting of an amorphous Ni-rich Ni-Si overlayer and a poly-crystalline NiSi₂ layer at the film/Si interface due to a supercooling of the melted region. Low fluence LA causes intermixing of Ni and Si atoms, but does not induce significant Ni silicidation. A post-LA RTA at 400 °C can be used to promote

further NiSi formation from the intermixed layer without causing any significant diffusion to the laser annealed Ni profile and B profile tail.

8.2 Recommendations for Further Research

Several research topics on LA should be studied for possible integration of LA in the semiconductor industry.

- 1) LA has received a tremendous interest in the formation of ultra-shallow p^+/n junctions due to difficulties in forming ultra-shallow p^+/n junctions using conventional processes. It has been found that LA produces highly activated ultra-shallow p^+/n junctions that meet the requirements for nano-scale pMOSFET. Further study on LA of ultra-shallow n^+/p junctions using arsenic or phosphorus species is also of great importance. Further study should aim at the simultaneous formation of both p-type and n-type source/drain extensions using a single LA process.
- 2) This project has shown that the combination of subamorphization implantation and non-melt LA produces ultra-shallow junctions with good activation and leakage characteristics. The improvement is attributed to a vacancy-rich surface region generated by the Si^+ subamorphization implantation. Further optimization of the subamorphization implantation in creating a more distinctive vacancy-rich surface region should be explored.
- 3) Atomistic modeling on the deactivation and enhanced diffusion incurred upon post-LA RTA provides a better understanding on the thermal stability of laser-activated dopant. Since LA produces ultra-shallow junctions with high dopant activation in a metastable state, post-LA thermal cycles may cause significant degradation to the LA induced

junctions. It is therefore important to have a good understanding about the interactions between the activated dopant atoms and the residual defects and to predict the final junction characteristics upon post-LA thermal cycles.

- 4) It has also been shown that LA can be used for the simultaneous formation of Ni silicide and highly activated ultra-shallow junctions (refer to chapter 7). The profiles of Ni silicide and dopant distribution are controlled by the LA conditions. This can be applied in advanced MOSFET fabrication that requires tailoring of silicide and dopant profiles, such as workfunction adjustment of fully-silicided (FUSI) gate and schottky source/drain MOSFET.

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Patent

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