LOW TEMPERATURE SOLID PHASE EPITAXIAL REGROWTH OF ION IMPLANTED BORON IN SILICON

By

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This dissertation is dedicated to the greatest man I know, my dad.
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The ever-changing nature of silicon-based microelectronic devices demands material properties be modified to meet device expectations. Moving to ultra shallow junction material requires high dopant activation while maintaining shallow electrical junctions. Ion implantation processes inherently produce lattice damage in silicon, bringing about the phenomenon of transient enhanced diffusion upon subsequent moderate temperature annealing. Solid phase epitaxial regrowth, on the other hand, will propagate at lower temperatures, thus minimizing diffusion while still activating dopants. It is the low temperature (500-650 °C) solid phase epitaxy process that is of main interest for the work contained within.

Several experiments are designed to further understand the solid phase epitaxy process and determine its usefulness for creating the desired material properties. Studies of ultra shallow junction, regrown materials examine the possibility of generating a material suitable for future technology. Applying theoretical predictions to experimental results
yields additional proof of satisfactory material character. Other work includes examination of junction leakage when end-of-range damage is adjusted relative to the boron implant tail. While placing end-of-range damage near the boron tail maximizes leakage, reducing the damage density helps lower leakage currents. This reduced leakage is also seen when the defect line length is minimized.

Characterizing ultra shallow junction materials is sometimes difficult so an experiment with deeper layers will provide more accurate results. Furthermore, this experiment investigates two different preamorphizing species to determine if there are any benefits to material characteristics. The sheet resistance results observe the tradeoffs between 100% activation of an implant and reaching a minimum sheet resistance. Activation of boron to levels of ~2-3x10^{20} \text{ cm}^{-3} is attained with silicon preamorphization and germanium preamorphized samples yield lower activation (~1-2x10^{20} \text{ cm}^{-3}). Here again theoretical predictions match experimental data. Sheet resistance results for germanium preamorphized materials reveal higher values than silicon counterparts due to increased scattering events from germanium impurities. This is reflected in a slightly decreased mobility of the germanium preamorphized samples. Altogether these experiments provide a better understanding of low temperature solid phase epitaxy methods and assist with future experimental designs.
CHAPTER 1
INTRODUCTION

The science and engineering of semiconductor materials have evolved some remarkable changes over the last several decades. The trends in semiconductor advances show little signs of slowing down due to the continuous demand for increased computing power. Therefore, the need for research directed at material design and enhancement calls for science to continue evolving. Innovations by scientists lead to these advances, but these innovations cannot be accomplished without an understanding of the work at hand. This chapter presents the motivation behind the research contained within and a brief description of the work and results of interest.

Motivation

The metal oxide semiconductor field effect transistor (MOSFET) has been a workhorse for the semiconductor industry for quite some time now. Due to the vast knowledge and understanding of the MOSFET along with its scaling capabilities it has become an important device for current technology. The durable design allows devices to be scaled into the submicron range, which in turn permits higher device densities and increased performance demanded by consumer markets. As scale down continues for future generations there are fundamental limitations that are being met. It is these limitations that call upon scientists to investigate problems and possible solutions to scaling down.

To begin understanding what is needed for scale down it is necessary to examine the MOSFET. Figure 1.1 shows a cross sectional view of a typical MOSFET. The ultimate
goal is to shorten the channel length \( (L_{\text{ch}}) \), thus making the device faster and permitting higher densities of devices. Simply shortening the channel length while leaving other dimensions constant harms the overall performance of the device. One of the main concerns is scaling the source-drain extension (SDE) depth, \( x_j \), while decreasing transistor size. If junction depths remain at deeper levels the lateral depletion under the gate is larger resulting in increased charge sharing leading to short channel effect (SCE). The large off-state leakage is detrimental to the device as it will always be in the “on” state. While it may be possible to decrease the junction depth, \( x_j \), at the same time it is also necessary to maintain a low sheet resistance, \( R_s \), in the implanted layer.

Having low material sheet resistance is crucial since the combination of millions of transistors can accumulate to very large resistances overall. Guidelines for the desired material properties for future generations of semiconductor devices are given in the International Technology Roadmap for Semiconductors (ITRS) [ITR01]. Figure 1.2 shows a plot of sheet resistance, \( R_s \), versus junction depth, \( x_j \), for results obtained by conventional implant and annealing techniques along with the requirements each year given by the ITRS. Clearly there is a need to reduce junction depths while maintaining relatively low \( R_s \). This is done by activating higher levels of dopants in silicon and using ultra low energy (ULE) ion implantation. Using the sheet resistances given in the ITRS and estimating the mobility of a uniformly doped layer, a plot of active concentration range required for future device generations is given in figure 1.3. It can be seen that active dopant levels need to reach near \( 10^{20} \text{ cm}^{-3} \), which is considerably high. Conventional implantation can introduce these levels of dopants, but activation will not
occur without further processing steps. It is the later processing steps in which active
dopant limitations appear.

Conventional processing relies on rapid thermal processing (RTP) to repair the
damage generated by the implantation step and activate dopants. The greatest limitations
occur for the dopant boron, which is known to achieve lower active concentration levels
and diffuse rapidly by a process known as transient enhanced diffusion (TED). With
both low levels of activation and rapid diffusion it becomes more difficult to reach future
low $R_s$ and small $x_j$ prescribed by the ITRS. Thus new avenues to activating dopants
with minimal diffusion are being explored.

Objectives

The research contained within this dissertation focuses on increasing active levels of
boron while minimizing diffusion in the tail end of implants. To do this, a technique
known as solid phase epitaxy (SPE) regrowth is employed. The foundation of this
method lies in reordering a doped amorphous layer back to crystalline silicon at low
temperatures (500–650 °C). During this solid-state recrystallization process the dopant is
activated and the diffusion is minimal due to the low temperatures used. Another
advantage to SPE regrowth is that dopant implants into amorphous layers reduce
channeling, ensuring shallower junction depths.

This work concentrates on several experimental approaches to examining the viability
of SPE as a means to future processing. Contributions to the scientific community will
include the following:

- Brief examination of how the SPE regrowth process works and the experimental
techniques used to characterize SPE.
• Evidence of the ability to form ultra shallow junction materials that are sufficiently active for device fabrication.

• Application of models to explain phenomena seen in SPE regrown material and extract attained active concentration levels.

• Discussions on the effect of junction depth relative to end-of-range damage and its impact on leakage current.

• Optimization of process conditions used for SPE regrown materials and the effect that different preamorphization species may have.
Figure 1.1. Schematic of a typical MOSFET device showing the gate length, $L_g$, and channel length, $L_{ch}$.
Figure 1.2. ITRS roadmap for sheet resistance requirements as a function of junction depth.
Figure 1.3. Active concentrations needed to meet technology roadmap by using sheet resistances and estimating mobility of active layer.
CHAPTER 2  
LITERATURE ASSESSMENT

This chapter presents an overview of some current knowledge involved in integrated circuit (IC) processing and physics of silicon-based materials. A review of ion implantation processing and its side effects is briefly discussed. Considerations extend into the consequences of annealing damage generated by ion implantation and attempt to explain the transient enhanced diffusion (TED) phenomena. Following TED details, a synopsis of the solid phase epitaxial (SPE) regrowth process and why it is employed is given. There have been a number of experiments observing the effects of substrate orientation and impurities on SPE regrowth, so results are summarized and displayed. Then regrowth models are imparted to account for the observations.

**Conventional Processing and Limitations**

Dopant introduction into silicon for IC manufacturing is done predominately by ion implantation. While implantation offers excellent control of impurity amounts it also produces damage to the silicon crystalline lattice. Upon implantation of the silicon surface with sufficiently heavy elements, electronic drag and nuclear collisions will slow the implanted species until it comes to rest within the silicon matrix. The final resting place may not be on a lattice site (thus an interstitial site) and nuclear collisions may leave other silicon lattice atoms displaced from their lattice sites. The displaced silicon atoms are known as interstitials while the vacant lattice sites are vacancies and together they are called Frenkel pairs. Although at any temperature there is an equilibrium number of interstitials and vacancies (also known as point defects), the ion implantation
will generate a super-saturation of point defects that needs to be annealed out to improve material quality. If the dose of the implant is high enough to create ~10-15% damage to the lattice, the crystalline structure will collapse to an amorphous material which is randomly ordered. Amorphous materials and highly damaged crystals with dopants sitting in interstitial sites yields poorly conducting material. Therefore, it is necessary to anneal the damage and force dopants onto substitutional sites.

**Annealing and TED**

Ridding the damage created by ion implantation is typically fulfilled by thermally heating the silicon. This gives the point defects enough energy to migrate such that vacancies and interstitials may combine to annihilate one another. Unfortunately the damage annealing is not as simple as recombination since there are extra interstitials introduced from the implant. It is the net extra interstitials that cause problems upon annealing.

Annealing of ion-implanted silicon at elevated temperatures can cause dopants to diffuse much faster than anticipated for equilibrium diffusion. It appears that boron is the most severely affected dopant so discussions will focus on boron. This fast diffusion is known as TED because it only occurs for a distinct period of time, and then equilibrium diffusion takes over again [Mic89, Pac90]. Models to explain this phenomenon describe how the excess interstitials from ion implantation are responsible for TED [Eag94, Eag95]. The silicon interstitials cluster together at which point the clusters either grow into \{311\} extended defects or dissolve and release interstitials to other, more stable clusters or extended defects. The silicon interstitials are also responsible for “kicking out” substitutional boron to interstitial positions allowing rapid diffusion of boron. The \{311\}s are somewhat stable entities but they will release silicon interstitials over time.
allowing further diffusion of boron. It is also possible that the \{311\} defects may unfault into dislocation loops under the right conditions. Loops are even more stable and may prolong the enhanced diffusion. The diffusion period is certainly controlled by the anneal temperature whereby the greater the temperature the shorter the transient period and diffusion [Cow99, Eag95]. Transient diffusion may last for hundreds of hours at 650 °C or on the order of seconds at 950 °C. An excellent review of boron TED in silicon is given by Jain et al. [Jai02].

With the advent of TED there is a major concern about forming ultra-shallow junctions that have sufficiently activated dopant levels. TED may drive the boron tail too deep to obtain acceptable MOSFET device characteristics. As boron seems to be the limiting dopant, there is much research into attaining high activation while minimizing diffusion. Recent studies show that using ultra-high ramp rates and high temperatures will minimize TED [Bou01, Jai02]. The other approach would use temperatures low enough that diffusion is negligible but still sufficient to activate dopants. SPE regrowth uses this approach to form highly active, shallow junction materials.

**Introduction to SPE Regrowth**

Solid phase epitaxial growth processes all involve the same general characteristics when talking about silicon, but there are several approaches that may be taken. In general the process involves an amorphous layer in intimate contact with a crystalline layer, which upon heating to sufficient temperatures allows the amorphous layer to crystallize using the crystalline layer as a seed. A well-defined amorphous/crystalline (α/c) interface moves toward the surface at a rate dependent on several factors such as orientation and impurity concentration. The amorphous layer may be deposited onto a crystalline layer or substrate via chemical vapor deposition (CVD) or similar technique.
The obvious disadvantage to using deposition techniques would be the need to highly control the amount of impurities on the surface of the substrate to ensure good quality growth later. A more common approach involves bombarding the surface of a crystalline substrate or layer with a heavy ion at sufficient dose to amorphize part of the crystalline material. Generally this is done using ion implantation, which has the advantage of producing a cleaner amorphous layer and α/c interface. If only part of the crystalline layer is amorphized, then the remaining crystalline layer can act as a seed to recrystallize the amorphous layer upon heat treatment. This approach is typically employed and is the basis for work in this dissertation.

Several steps are involved with the production and recrystallization of amorphous layers used in the work here. As already alluded to, the amorphous layers are generated by ion implanting the surface of silicon wafers with a heavy ion such as Si⁺ or Ge⁺. Care is taken to choose an appropriate tilt and twist to minimize the channeling of ions, and implant energies are typically chosen to amorphize a layer from the surface down to some desired thickness. It is also possible for a high-energy implant to create a buried amorphous layer that would regrow from two directions. After amorphizing another impurity may be introduced into the material to examine how the recrystallizing layer behaves. Thermal heating [Nar82e, Nar83, Poa84, Wil83c, Wil83d], electron beam heating [Tim85, Tim86], ion beam assisted regrowth [Ell87a, Ell87b, Ker86, Wil85], and laser heating [Kok82, Ols84, Ols85a, Ols85b] techniques aid in regrowing the amorphous layer by giving the system enough energy for reorganization. Upon recrystallization, impurities may become trapped onto substitutional lattice sites allowing metastable conditions to be met. Typical SPE regrowth begins to occur at temperatures as low as
~400-450 °C [Poa84, Wil83c, Wil83d] on up to temperatures below the melting point of amorphous silicon with regrowth rates dramatically increasing for increasing temperature. Ion beam assisted techniques have shown it is possible to regrow amorphous layers at temperatures >170 °C while lower temperatures may actually increase the thickness of the amorphous layer [Ell87a, Ell87b, Ker86]. It is particularly interesting to examine temperatures ≤650 °C where diffusion is minimal. However there are a number of things that can affect how regrowth proceeds.

**Substrate Orientation**

Early work in ion implantation of silicon and germanium shows amorphous layers reorder at significantly low temperatures [May68]. In general SPE regrowth velocity (the velocity of the α/c interface) follows an Arrhenius temperature dependence of the form:

\[
v = v_o e^{\left(\frac{E_a}{kT}\right)}
\]

where \(v\) is the regrowth velocity, \(v_o\) is the pre-exponential factor, \(E_a\) is the activation energy, \(k\) is Boltzmann’s constant, and \(T\) is temperature in K. Csepregi et al. [Cse75, Cse78] notice this and set up a series of experiments that involve examining the effect of SPE regrowth on variously oriented slices of silicon between 450 and 575 °C. Their work shows regrowth has an activation energy of about 2.35 eV for all orientations. They also find <100> silicon regrows at a rate about 3 times faster than <110> silicon and about 25 times faster than <111> silicon. Drozd and Washburn [Dro82] find the same trends for regrowth at different orientations but obtained activation energies of 2.9 eV and regrowth rates that are 2.3 and 20 times slower for <110> and <111> oriented silicon respectively. However, Drozd and Washburn seem to neglect the effects of using P+ as an amorphizing species on the regrowth. Lietoila et al. have repeated regrowth of
intrinsic <100> silicon and find the activation energy to be 2.85 eV and pre-exponential factor of 2.11x10^{19} Å/min [Lie82]. They neglect the initial stages of regrowth in their measurements because the tail end of the amorphous layer regrows faster, which they attribute to incomplete amorphization.

Later experiments made use of more sophisticated techniques for heating and measuring regrowth to obtain more accurate results over wider temperature ranges. The work by Licoppe and Nissim [Lic86] examines a larger temperature range (500-900 °C) using furnace and laser heating while measurements are acquired from time resolved reflectivity (TRR) techniques. Their findings produce values of 3x10^8 cm/sec and 2.7 eV for \( v_0 \) and \( E_a \) respectively [Lic86]. Olson uses laser heating and TRR to discover values of 3.07x10^8 cm/sec and 2.68 eV for \( v_0 \) and \( E_a \) respectively [Ols85a]. The research group including Park et al. and Jeon et al. employ vacuum annealing and laser interferometry for regrowth measurements [Jeo89, Jeo90, Par88]. Values for \( v_0 \) vary between 1.5–3.5x10^{18} Å/min while \( E_a \) is between 2.68 and 2.74 eV over all the experiments performed. Thus there seems to be some variation in the reported activation energies and pre-exponential factors.

Distinct differences in <111> or near <111> oriented regrown silicon is observed as compared to other orientations [Cse75, Cse78, Dro82]. If the orientation of silicon is within \(~16^\circ\) from the <111> axis a larger number of residual defects and twins will occur [Cse78]. The defects may be an artifact from how the regrowth progresses and will be discussed further when regrowth models are presented. The <111> oriented silicon has also been observed to exhibit an initial slow regrowth followed by a 2.5-3 increase in rate, which has been attributed to the formation of the twins [Cse78, Dro82].
**Impurity Effects**

Introducing impurities into an amorphous layer that is to be recrystallized will have a profound impact on regrowth characteristics. Each species as well as its concentration can vary the behavior observed. Here the effects of various impurities on regrowth are summarized by type. Many of the studies are done at $\leq 650 \, ^{\circ}C$ and it will be assumed the results discussed below are low temperature unless otherwise specified. Much of the early work also relies on Rutherford backscattering (RBS) measurements to examine materials and estimate the substitutional fraction of incorporated impurities; thus care needs to be taken when interpreting the reported substitutional concentrations as the species may be on or only near lattice sites. The most useful doping impurities, boron (B), phosphorus (P), and arsenic (As) are discussed first followed by lesser-exploited group III and V dopants such as aluminum (Al), gallium (Ga), indium (In), nitrogen (N), antimony (Sb), and bismuth (Bi). A brief review of group IV elements (excluding silicon) is also given. Then some comments on the effect of noble gasses, various group VI and halogens, and some transition metals are made for completeness. Finally observations of co-implanting group III and V elements are described to examine compensation effects.

**Boron**

The initial work involving the impact of B implanted into $\langle 100 \rangle$ silicon on regrowth shows regrowth velocity is greatly enhanced by moderate to high concentrations of B [Cse77]. Csepregi et al. found boron concentrations of $\sim 2.5 \times 10^{20} \, \text{cm}^{-3}$ will increase the regrowth rate 25 times that of intrinsic silicon and concentrations of $10^{21} \, \text{cm}^{-3}$ will increase the rate $\sim 10$ times [Cse77]. This enhancement in regrowth is larger than any of the other impurity species effects studied. It is also shown in their work that the
activation energy is considerably lower (1.9 eV) which may help explain the enhancement. The higher B concentration regrowth also shows greater numbers of residual defects exist and could reduce regrowth rate. One of the main flaws with this work is regrowth is measured over a changing concentration profile due to the Gaussian nature of ion implantation.

Similar studies of B effects on regrowth are performed by Suni et al. They produce a more uniform concentration of B through multiple implants and achieve a concentration of ~2x10^{20} cm^{-3} [Sun82a, Sun82b, Sun82c]. The regrowth temperature is limited to 475 and 500 °C but they find that the enhanced regrowth rate is about 15-17 times that of intrinsic silicon. However intrinsic regrowth is not measured in the experiment and is compared to the work of Csepregi. A similar experimental approach taken by Ho et al. [Ho84] uses multiple energy implants to create fairly uniform doping concentrations for measurements. Their work also looks at how substrate orientation changes enhancement but only over the 500-550 °C temperature range. It is found that an enhancement factor of 20.8 times the rate of intrinsic <100> silicon is achieved while enhancements of only 13.3 and 11.0 are seen for <110> and <111> silicon respectively. They also find there is some beneficial influence that doping has on the quality of regrowth by reducing the amount of residual damage in the regrown layer. Kerkow et al. [Ker84] study enhancements at 515 °C versus B concentration, finding the greatest regrowth rate occurs at ~4x10^{20} cm^{-3} and then rapidly falls as concentration increases. They also find concentrations above ~10^{21} cm^{-3} cause the regrowth rate to fall below that of intrinsic silicon.
Since regrowth observes an Arrhenius dependence, the time scale to complete recrystallization of an amorphous layer may become very small at temperatures >650 °C. This limits experimental observations to lower temperatures unless in-situ measurements and very high temperature ramp rate equipment is used. Timans et al. [Tim85] accomplish this by using electron beam heating and TRR measurements for temperatures >700 °C. They also find enhancements of B doped material and are able to determine the regrowth rate change as the α/c interface moves through changing B concentration levels. Olson et al. use a cw laser for heating to higher temperatures and take data using TRR measurements [Ols85a, Ols85b]. From their experiments of a 10 keV, 3x10^{15} \, \text{cm}^{-2} \, \text{B}^+ implant, v_0 is measured to be 5.35x10^8 \, \text{cm/sec} and E_a is 2.59 \, \text{eV}, which is slightly less than intrinsic silicon [Ols85a, Ols85b]. One of the drawbacks to using TRR is the need for models to fit the experimental data, especially when interface effects become significant. Park et al. study regrowth of B implanted layers in the concentration range of 2x10^{18}-5x10^{20} \, \text{cm}^{-3} fitting their data by normalizing velocity by the intrinsic rate and normalizing the B concentration by intrinsic carrier concentration [Par88]. Essentially they find a linear correlation between regrowth velocity and B concentration except at the highest concentrations.

Other phenomena associated with the regrowth of B doped amorphous layers are observed as well. At typical regrowth temperatures (≤650 °C) there tends to be very little motion and redistribution of B. There is some evidence, however, that shallow layers may exhibit slight motion [Hua96] and B may pile up at the end-of-range (EOR) damage [Lin02]. Elliman et al. show the boron motion is due to diffusion in the amorphous phase for experiments at 600 °C [Ell98]. Olson et al. find regrowth at high temperatures will
not induce the formation of polycrystalline surfaces at high boron concentrations but the addition of fluorine (F) (from BF$_2^+$ implants) will cause a competition between SPE and random crystallization [Ols85b]. Finally, Tsai and Streetman find amorphous layers containing >10$^{20}$ cm$^{-3}$ B may regrow but active levels reach ~10$^{20}$ cm$^{-3}$ [Tsa79a].

**Phosphorus**

In the early studies of P effects, it is noticed that regrowth is enhanced similarly to B [Cse77]. Reports on <100> and <111> silicon indicate regrowth rate increases with increasing P concentration until concentrations >2x10$^{20}$ cm$^{-3}$ are reached, at which point regrowth decreases from a factor of 6 enhancement [Cse77]. The activation for this work is 2.35 eV, which matches their intrinsic silicon activation energy. Like their work for boron, the regrowth is performed using an implanted P profile, which has varied concentration with depth.

Results by Licoppe and Nissim are also gathered from samples implanted at various doses. In their work a mean concentration in the implanted profile is determined for each of the doses. They find the average activation energy actually drops slightly to 2.5 eV (compared to 2.7 eV for their intrinsic silicon) and the pre-exponential term varies between 3x10$^7$ and 3x10$^9$ cm/sec [Lic86]. By measuring the activation at different P doses they find the activation energy starts at 2.7 eV and drops to ~2.38 eV, leveling off for P concentrations of ~4x10$^{20}$ cm$^{-3}$ and higher. Kerkow et al. look at a range of P concentrations and find a maximum enhancement at ~4x10$^{20}$ cm$^{-3}$ followed by a rapid drop in regrowth rate as concentrations increase [Ker84]. They find P concentrations >10$^{21}$ cm$^{-3}$ are needed to slow the regrowth rate below the intrinsic rate, which is similar to B.
Results from TRR and cw laser interferometry experiments confirm the findings on P enhanced regrowth. Timans et al. measure regrowth rate as a function of the P concentration using the implant profile of P [Tim85]. Measurements at temperatures $>700 \, ^\circ\text{C}$ show similar trends that increased P levels increase the regrowth rate. Jeon et al. also track the P enhancement but at lower temperatures ($460-590 \, ^\circ\text{C}$) [Jeo89]. They extract out activation energies that start at 2.65 eV for intrinsic silicon and decrease quickly to 2.45 eV at $\sim 4 \times 10^{20} \, \text{cm}^{-3}$ P while $v_o$ jumps between $2 \times 10^{17}$ and $1.2 \times 10^{18} \, \text{Å/min}$. Another interesting discovery they make is the enhancement starts at $\sim 10.7$ times intrinsic silicon at low temperatures but drops linearly with increasing temperature to $\sim 6$ at 590 $^\circ\text{C}$.

A more careful approach to creating constant doping concentration is taken by several authors [Ho84, Lie82]. Both experiments use multiple implants to generate a distance of constant P concentration where measurements are taken. Enhancements of P layers doped to $\sim 2 \times 10^{20} \, \text{cm}^{-3}$ are 8.0, 7.2, and 9.2 for $\langle 100 \rangle$, $\langle 110 \rangle$, and $\langle 111 \rangle$ silicon respectively [Ho84]. Essentially there is little change in the enhancement factor for each orientation. Experiments by Lietoila et al. find a fairly constant doping of $1.7 \times 10^{20} \, \text{cm}^{-3}$ produces a $v_o$ of $6.82 \times 10^{17} \, \text{Å/min}$ and an activation energy of 2.5 eV, which is 0.35 eV below their intrinsic value [Lie82]. Overall it seems reasonable to assume there will be a slight drop in activation energy when doping with moderate to high P levels.

Final comments on the motion of P during regrowth are made. Sadana et al. implanted $\langle 111 \rangle$ silicon with P$^+$ at levels sufficient to create buried amorphous layers [Sad82]. During regrowth it is found P did not redistribute for moderate concentrations and
temperatures of 650 °C and below. But as has already been alluded to, sufficiently high concentrations of P will inhibit or halt regrowth completely.

**Arsenic**

As one might expect by now, the presence of As will increase the regrowth rate with increasing As concentration until a critical concentration is reached. At this point the regrowth rate begins to decrease with increasing As concentrations similarly to the cases of B and P. Early measurements [Cse77] show As behaves very similarly to P in that it enhances regrowth by a factor of 6 for concentrations of $2.4 \times 10^{20}$ cm$^{-3}$. As before though, the enhancement is averaged from ion implanted samples that have varying As concentrations with depth.

Suni et al. use multiple implants to create a more uniform As concentrations of $\sim 10^{20}$ cm$^{-3}$ [Sun82a, Sun82b, Sun82c]. At this concentration the recrystallization rate rose to a value of $\sim 4-5$ times the intrinsic rate. Kerkow et al. examine the regrowth rate at 515 °C as a function of As concentration and determine the maximum rate is at $\sim 4 \times 10^{20}$ cm$^{-3}$ and begins to slow down at higher concentrations [Ker83, Ker84]. With increases over $2 \times 10^{21}$ cm$^{-3}$ the rate of regrowth begins to fall below the intrinsic rate.

The in-situ TRR and cw laser interferometry experiments also show the same effects of As. The >700 °C experiments by Timans et al. [Tim85] verify the enhancement and so do measurements in the temperature range of 500-900 °C [Lic86]. Licoppe and Nissim conclude the activation energy will drop from $\sim 2.7$ eV for intrinsic silicon to $\sim 2.36$ eV at concentrations of $8 \times 10^{19}$ cm$^{-3}$ and level off at higher concentrations [Lic86]. The pre-exponential factor varies from $10^{7}$ to $10^{8}$ cm/sec. The experiments of Jeon et al. [Jeo90] measure a drop in activation energy with As concentration also. Intrinsic silicon measures $E_a$ of $\sim 2.65$ eV and drops quickly to $\sim 2.47$ eV at $\sim 1.3 \times 10^{20}$ cm$^{-3}$ As
concentration. From there they show $E_a$ actually increases with further increasing As levels. The $v_o$ varies among $2 \times 10^{17}$ and $1.2 \times 10^{18}$ Å/min. Finally, Kokorowski et al. measure regrowth of As-implanted amorphous layers utilizing furnace and laser annealing with TRR measurements in the 750-1550 K temperature range [Kok82]. Low concentration As ($1.9 \times 10^{19}$ cm$^{-3}$) forces $v_o$ to $5.75 \times 10^7$ cm/sec and $E_a$ to $2.62$ eV while higher concentration ($4.3 \times 10^{19}$ cm$^{-3}$) has $v_o$ of $2.21 \times 10^7$ cm/sec and $E_a$ of $2.52$ eV.

Electrical activity of As is also shown to exceed solid solubility limits in SPE regrown materials. Nishi et al. performs early work using As implants at $1 \times 10^{15}$ cm$^{-2}$ dose to amorphize the surface and study regrowth in <100>, <110>, and <111> silicon [Nis78]. Using Hall and stripping Hall measurement techniques, they show As will almost completely activate for <100> orientations while about 80 and 65% of the As activates in regrown <110> and <111> silicon respectively. Active As concentrations in excess of $10^{20}$ cm$^{-3}$ are accomplished in <110> silicon, found from stripping Hall. They also find an average $E_a$ of $2.7$ eV for all orientations of silicon, however they assume the As concentrations have no effect on $E_a$. Work by Lietoila et al. also shows As concentrations of up to $5 \times 10^{20}$ cm$^{-3}$ are possibly achieved but further annealing at 560 °C will deactivate As [Lie81]. This shows As does exhibit metastable conditions and will deactivate at low temperatures.

Further studies on the effects of As on the kinetics of regrowth demonstrate interesting behavior. Olson et al. run an experiment varying As dose and higher ($\geq 640$ °C) regrowth temperature to examine the effect of the SPE rate [Ols84]. The findings include observations of increasing regrowth rate as concentration increases to $\sim 4-5 \times 10^{20}$ cm$^{-3}$ As at which point the rate begins to slow for higher concentrations. The Gaussian nature of
implants allows moderate concentrations deep in the substrate as well as near the surface. As the $\alpha/c$ interface passes through the first region of moderate As concentration regrowth is accelerated followed by retardation at the peak of the implant. When the interface reaches the moderate concentrations nearer to the surface the regrowth rate is not as enhanced as earlier. Olson et al. also find increasing the anneal temperature results in less retardation at the peak concentrations. In later work [Ols85b] it is observed that temperatures in excess of 980 °C induce a competition of polycrystalline growth with SPE growth. Polycrystalline material may form in the amorphous material ahead of the $\alpha/c$ interface if sufficiently high As concentrations ($>4\times10^{20}$ cm$^{-3}$) are present. A final comment on the work by England et al. [Eng90] is how regrowth of amorphized polysilicon proceeds with the presence of As. It is shown smaller grains slow the regrowth rate of polysilicon and create rough $\alpha/c$ interfaces. The presence of As not only increases regrowth rates but also smoothens the $\alpha/c$ interface.

**Group III Elements Excluding Boron**

In this section a review of group III element’s (aluminum (Al), gallium (Ga), indium (In), and thallium (Tl)) effect on SPE regrowth is offered. In general it is seen that these elements also increase the regrowth rate, but not as much, and maximum incorporation limits are usually smaller. These elements also have the disadvantage of being very mobile upon recrystallization if concentrations become too high. These and other effects observed by various authors are presented.

Despite its lower volume of use in the electronics industry, there is a fair amount of SPE studies done using In. Narayan and co-workers have a few studies [Fle81, Nar82a, Nar82b, Nar82c] and several reviews [Nar82d, Nar82e, Nar83] that cover the specifics of low temperature ($<600$ °C) SPE regrowth with In implants. They find the In is capable of
enhancing the regrowth rate slightly, but too much In retards regrowth similar to other dopants. Indium will sit on substitutional sites upon regrowth if the implant dose (directly related to the concentration) is sufficiently low. At higher concentrations the In is rejected and “pushed out” of the crystalline silicon ahead of the α/c interface and large accumulations at the surface may occur. Precipitation of In is also possible in the crystalline and amorphous materials during regrowth. From their RBS measurements, Narayan and co-workers find a maximum substitutional/near substitutional concentration of $5.5 \times 10^{19}$ cm$^{-3}$ and determine it to be ~2 orders of magnitude higher than equilibrium values. Narayan also does a number of transmission electron microscopy (TEM) experiments and find high In concentrations may lead to twin formations and the onset of polycrystalline regrowth. It is also shown that higher anneal temperatures allow In to diffuse and decorate loops in the EOR and eventually redistribute back to equilibrium levels.

Williams and co-workers perform several $<650$ °C studies on In implanted silicon as well [Nyg87, Wil81, Wil82a, Wil82b, Wil83b]. Basically they find the same results concerning the behavior of In in silicon during SPE regrowth. Maximum concentrations of $\sim 5 \times 10^{19}$ cm$^{-3}$ are observed along with the “push out” effect at higher concentrations. They claim a dose of $\sim 10^{15}$ cm$^{-2}$ In is sufficient for the onset of the “push out” phenomena but do not specify an energy of the implant used. Later work shows $>0.5$ at% In will induce polycrystalline regrowth [Nyg87]. The polycrystalline formation allows rapid diffusion of In, probably along grain boundaries.

Several other studies involving In confirm similar results and note some other interesting observations. Kerkow et al. did studies of regrowth at 515 °C finding the
greatest regrowth enhancement at $\sim 5-6 \times 10^{19} \text{ cm}^{-3}$ In concentration [Ker84]. Blood et al. examine lower dose In implants that do not redistribute at low temperatures [Blo79]. They find low temperature annealing allows retention of dose but subsequent higher temperature regrowth anneals cause loss of In. Estimates of substitutional fraction suggest only $\sim 40\%$ or less of In will sit on lattice sites with decreases as anneal temperature rises. The study uses $\langle 100 \rangle$, $\langle 110 \rangle$, and $\langle 111 \rangle$ oriented substrates and findings include the $\langle 100 \rangle$ silicon retains more In than $\langle 110 \rangle$ which retains more than $\langle 111 \rangle$ following high temperature treatment.

The work of Narayan and co-workers also examines the effect Ga has on SPE regrowth. Essentially, Ga will behave very similar to that of In [Nar82a, Nar82d, Nar82e, Nar83]. Low concentrations allow enhanced, complete regrowth and almost total incorporation while higher concentrations inhibit regrowth and may induce polycrystalline formation. Redistribution of Ga is possible by similar “push out” effects as well as precipitation formation. The maximum substitutional concentration is estimated to be $2.5 \times 10^{20} \text{ cm}^{-3}$, which is about an order of magnitude higher than equilibrium. Kerkow et al. find the maximum regrowth enhancement at concentrations of $\sim 2-3 \times 10^{20} \text{ cm}^{-3}$ Ga at 515 °C [Ker84].

Some final comments on the remaining group III elements (Al and Tl) are made to finish this section. Kerkow et al. examines Al, finding it to be very similar to the case of Ga. The maximum regrowth enhancement occurs at $\sim 2-3 \times 10^{20} \text{ cm}^{-3}$, however Al will redistribute at low temperatures ($<500 \degree \text{C}$) which would make it difficult to control doping profiles [Ker84]. Williams and Short study Tl determining a maximum
metastable substitutional concentration of $4 \times 10^{19}$ cm$^{-3}$, about 2 orders of magnitude higher than equilibrium [Wil82b]. They also find “push out” begins at $\sim 10^{15}$ cm$^{-2}$ dose.

**Group V Elements Excluding Phosphorus and Arsenic**

The remaining group V elements (N, Sb, and Bi) exhibit some varied effects on the SPE regrowth. In general, higher concentrations may lead to redistribution but the effect on regrowth rate is different for each species at low and high concentrations. Incorporation on to or near lattice sites is possible during SPE regrowth but is again reliant on the species.

Josquin and Tamminga present some of the early work on SPE of Sb implants. They examine high dose implants and the consequence of different implant temperatures on substitutionality and electrical activity. Overall at 600 °C it is found higher substitutional fractions are obtained for $<100>$ silicon versus $<111>$ silicon and electrical activity is substantially lower than estimated substitutional fraction [Jos77]. Implants at room temperature and below show about the same results while 700 °C implants are significantly less substitutional. They also find much lower substitutionality and electrical activity for higher temperature anneals.

The same groups who looked at Ga and In do many studies and reviews of Sb implanted regrowth. The efforts of Narayan and co-workers included Sb in their studies [Fle81, Nar82a, Nar82b, Nar82c, Nar82d, Nar82e, Nar83, Pen84]. Their studies include various doses of Sb and different implant temperatures. In general they find SPE regrowth proceeds as normal with some enhancement due to the presence of Sb. At higher concentrations ($>10^{21}$ cm$^{-3}$) regrowth rate begins to retard and may be halted without the formation of a polycrystalline layer. However, at higher anneal temperatures the growth of polycrystalline material may persist. Twin and precipitate formation at
higher dose implants is also discovered. Maximum substitutional/near substitutional concentrations obtained show levels of $2.5 \times 10^{21}$ cm$^{-3}$ are possible which is observed to be almost 3 orders of magnitude higher than equilibrium limits.

Numerous studies by Williams and co-workers find similar results on the characteristics of Sb in SPE grown material [Pog83, Wil81, Wil82a, Wil82b, Wil82c, Wil83b]. Metastable Sb concentrations of $1.3 \times 10^{21}$ cm$^{-3}$ are seen with an onset of the “push out” effect occurring at a dose of $\sim 7 \times 10^{15}$ cm$^{-2}$. Electrical characterization indicates concentrations of $3 \times 10^{20}$ cm$^{-3}$ become active [Wil82c] and active doses are considerably less than near substitutional doses [Pog83]. Damage analysis by Pogany et al. also shows low temperature (<650 °C) anneals create small defects at low concentrations while higher temperatures (>650 °C) begin to form larger loops [Pog83].

Several other experiments involving SPE regrowth of Sb implanted layers provide additional information. Kerkow et al. determine Sb concentrations between $10^{20}-10^{21}$ cm$^{-3}$ enhance the regrowth rate the greatest at 515 °C and higher concentrations begin to lower regrowth velocity [Ker84]. As concentrations increase the SPE regrowth rate drops quickly and will reach values well below the intrinsic rate. Other work with low dose Sb implants shows retention of Sb upon low and high temperature annealing is about 100% [Blo79]. It is also determined that fractional substitution is high for low temperature anneals but drops for higher temperatures. The substrate orientation has little effect except that <100> silicon has higher substitutional fractions after high temperature anneals.

Bismuth studies reveal somewhat similar trends in regrowth. Work using RBS shows maximum substitutional/near substitutional concentrations may reach as high as $4.5 \times 10^{20}$
cm$^3$, almost 3 orders of magnitude higher than equilibrium [Nar82d, Nar82e, Nar83]. Below this concentration SPE regrowth proceeds normally with total incorporation of Bi, while above this limit redistribution and defect formation occur. Twins and Bi precipitates are the result of high Bi concentrations and a non-planar a/c interface is observed [Nar82e]. Campisano et al. find $\leq 700 \, ^\circ C$ anneals produce implant layers with almost total incorporation of Bi for doses below $\sim 10^{15} \, \text{cm}^{-2}$ [Cam80b]. Higher doses appear to leave Bi off lattice sites. Finally, metastable Bi concentrations of $9 \times 10^{19} \, \text{cm}^{-3}$ are achieved by Williams and Short [Wil82b] and a dose of $\sim 2 \times 10^{15} \, \text{cm}^{-2}$ initiates “push out”.

Concluding remarks are made on nitrogen’s consequence on SPE regrowth at low temperature. Introducing even small amounts of N will reduce the regrowth rate of silicon [Ken77]. Increasing the N concentration only produces further retardation.

**Group IV Elements Excluding Silicon**

The group IV elements tend to behave similarly to silicon when introduced during SPE regrowth. At low concentrations the regrowth rates follow that of intrinsic silicon except for the case of lead (Pb). At higher concentrations, however, the impurities will greatly reduce the regrowth velocity. Brief descriptions of carbon (C), germanium (Ge), tin (Sn), and Pb follow for an introduction of their effects.

The most important preamorphizing species other than silicon is probably Ge. Since Ge is a heavier ion, lower doses will amorphize the surface of silicon leading to lower concentrations of Ge. Concentrations below $\sim 5 \times 10^{20} \, \text{cm}^{-3}$ Ge are shown to have very little impact on the SPE growth rate [Ker84]. Above this concentration the growth rate will slowly begin to decrease below the intrinsic value. Lee et al. observe decreases in regrowth rate but attribute it to defects generated due to the strain induced from high
concentrations of Ge [Lee93]. Other work confirms the presence of defects, which increase with increasing Ge concentrations and reduce the crystalline quality [Yam01]. Obviously due to the lattice mismatch between Ge and silicon, low concentrations of Ge are required to obtain quality regrowth.

Regrowth of Sn implanted silicon proceeds normally for low Sn concentrations while moderate to high concentrations slow regrowth. At concentrations below about 1 at% the regrowth kinetics are similar to intrinsic silicon, but above this level regrowth slows and polycrystalline layers may form [Tho89, Tho90]. The results of Thornton et al. indicate redistribution of high Sn concentrations during regrowth in both <100> and <111> substrates [Tho89, Tho90]. It also appears Sn helps stabilize the regrowth on <111> substrates by reducing twins and defects.

The effect of Pb on silicon regrowth is studied by Williams and co-workers [Wil77, Wil80a, Wil80b, Wil81, Wil82b]. Low concentrations of Pb allow regrowth to occur but retard the kinetics such that concentrations of up to ~7.5x10^{19} \text{ cm}^{-3} Pb may become substitutional. At higher concentrations the material turns polycrystalline and allows the redistribution of Pb. The onset of “push out” is ~10^{15} \text{ cm}^{-2} according to Williams and Short [Wil82b].

The final group IV element investigated is carbon. Concentrations below ~10^{20} \text{ cm}^{-3} show little effect on regrowth rate while higher concentrations slow growth [Ker84]. Kennedy et al. find concentrations of ~1.8x10^{20} \text{ cm}^{-3} C reduce the regrowth rate to about half that of intrinsic silicon [Ken77]. Overall it seems group IV elements behave similar to silicon in low concentrations with the exception of Pb.
Group VI, Halogens, and Noble Gasses

Included is a short discussion of various other group VI to VIII element effects on regrowth as a guide for potential contamination problems. Some of the elements may be mistakenly introduced during typical IC processing while others could be intentionally introduced. The elements included in the synopsis here are oxygen (O), tellurium (Te), fluorine (F), chlorine (Cl), neon (Ne), argon (Ar), and krypton (Kr).

The overview begins with the halogens, F and Cl, with particular attention to F, as it is introduced during IC manufacturing, usually in the form of BF$_2$. Small amounts of F will inhibit the SPE regrowth rate and completely halt regrowth if concentrations exceed $\sim 3 \times 10^{19}$ cm$^{-3}$ [Sun84]. While Sun et al. show F significantly lowers regrowth velocity it is also shown the addition of B with F will increase the velocity but still below that of intrinsic silicon [Sun84]. They also find F redistribution occurs during regrowth, even at temperatures as low as 500 °C. Tsai et al. also see redistribution in earlier work [Tsai79b]. This group also finds the F prefers to segregate ahead of the $\alpha$/c interface and at the EOR damage. Olson determines $v_o$ and $E_a$ for an 18 keV, $6 \times 10^{15}$ cm$^{-2}$ F$^+$ implant as $4.35 \times 10^9$ cm/sec and 3.06 eV respectively [Ols85a]. Similar results are found for the effects of Cl on regrowth. Chlorine tends to inhibit the regrowth more than F and adding B to increase regrowth velocity does not benefit regrowth as much [Sun84]. However, the redistribution of Cl is much less than that of F.

The noble gasses may be used as an inert carrier for annealing or etching so it is advantageous to look at the effect they have on regrowth. In general, results provide evidence of noble gasses reducing the regrowth rate of silicon with the possibility of halting regrowth at moderate concentrations [Ken77, Rev78, Wit78]. Defects are observed in as-implanted layers as well as regrown layers and may be in the form of
bubbles within the material [Cul78]. Low temperature regrowth redistributes impurities such as Ne but heavier impurities will not move as much unless higher temperatures are used [Res78, Wit78]. There is also the possibility of forming polycrystalline layers if impurity concentrations become too large.

The final elements under review are group VI, O and Te. Oxygen is a common impurity possibly introduced during ion implantation. The work by Kennedy et al. shows O has negative effects on regrowth [Ken77]. Increasing the O concentration only reduces the regrowth rate and it is estimated that O raises the activation energy of SPE regrowth [Ken77]. Therefore it seems beneficial to prevent the introduction of O into materials that are to be regrown. Other limited studies examine Te as an element that does not redistribute at low temperature regrowth [Cam80a]. There is not much else mentioned on Te, but it is unlikely to become a popular impurity used in silicon.

**Various Transition Metals**

To complete the discussion of specific element effects on regrowth, an assessment of select transition metals is undertaken. Studied species include silver (Ag), gold (Au), and copper (Cu). In all cases it is shown these metals are fast diffuser and will thus redistribute upon SPE regrowth [Cam80a, Jac87]. Redistribution occurs by “push out” of the impurities ahead of the α/c interface. It is also discovered the presence of Au may increase the regrowth rate of silicon [Jac87]. These metals are usually avoided as an impurity in silicon.

**Compensated Materials**

Compensated materials are the final topic for which experimental observations are discussed. What is meant by compensated material is semiconductor materials containing equal concentrations of group III and V impurities. Typical compensated
materials consist of B, Ga, or In along with P, As, or Sb. The study of these compensated materials is important because the effects need to be considered when a model for regrowth is hypothesized.

Experiments by Suni et al. look at the effect of intrinsic as well as B, P, As, and P+As implants with compensated regions where B concentrations are similar to either P or As concentration exhibit regrowth rates very close to intrinsic rates [Sun82a, Sun82b, Sun82c]. Temperatures of 475 and 500 °C are used. Lietoila et al. run similar experiments using multiple implants to create fairly uniform concentration layers and anneal in the 477-576 °C temperature range [Lie82]. They use B and P implants to dope to \( \sim 2 \times 10^{20} \text{ cm}^{-3} \) and determine activation energies of 2.8 eV and \( v_o=1.67 \times 10^{19} \text{ Å/min} \) for compensated material as compared to 2.85 eV and \( 2.11 \times 10^{19} \text{ Å/min} \) for intrinsic silicon. Timans et al. compare B/P and B/As compensated implants in their TRR study at temperatures >700 °C and also verify regrowth rates close to that of intrinsic silicon [Tim85]. Studies using In and Sb for compensating implants also show regrowth rates close to intrinsic silicon [Tho89, Wil83b]. Finally, Ga and As compensated implants are used by Kerkow et al. showing concentrations below \( \sim 10^{20} \text{ cm}^{-3} \) leave the regrowth rate close to intrinsic but above this the rate begins to decrease [Ker84].

After examining the effect different conditions have on SPE regrowth rate it is clear that a model for understanding regrowth may be complicated. Clearly the substrate orientation has a significant impact on regrowth with orientations near \(<111>\) involving complicated physics. Generally the common group III and V dopants increase regrowth velocity and the group IV elements have little impact on velocity until moderate to high concentrations. Various other impurities affect regrowth differently depending on
species while compensated implants have shown regrowth similar to intrinsic silicon. A summary of observed substitutional impurity concentrations by various authors is reported in table 2.1. The direct correlation of substitutional impurity concentration (from RBS) to impurity size is illustrated in figure 2.1. Plotting versus the absolute value of the difference in the tetrahedral bonding radius of silicon and impurity demonstrates elements with radii close to silicon are more readily incorporated onto lattice sites. Considering SPE regrowth is a very impurity species and orientation specific process it becomes obvious that all these things need to be considered in models.

Models for Solid Phase Epitaxial Regrowth

Despite the complexity of SPE, several mechanisms for the growth process are proposed to explain observed phenomena. Early accounts of SPE regrowth are either unable to explain effects or did not consider other effects observed in later experiments. There is still debate over whether the regrowth is controlled by defect diffusion to the α/c interface or reaction kinetically controlled. Here an overview of several proposed models is provided to introduce the reader to possible explanations of SPE growth.

Impurity Free Growth

To begin looking at models one needs to look at the thermodynamics of the system. For atoms to make the transition from amorphous to crystalline phase there must be negative net free energy change if indeed the crystallization process occurs. Making the transition requires the atoms to overcome a barrier before residing onto a lattice site. This is illustrated in figure 2.2 where $E_a$ is the activation energy (the barrier to overcome) and $\Delta G$ is the net free energy change. This is where the Arrhenius relationship for regrowth comes into play. The general form is given by 2.1 and extensions to this equation are shown for various authors.
Lau describes various approaches and inaccuracies of early models in his review [Lau78]. Csepregi et al. envisions regrowth as a bond-breaking mechanism of atoms in amorphous phase transforming to the crystalline phase [Cse78]. Growth occurs along \{111\} planes while growth in the \langle111\rangle directions is zero. It is also speculated that growth in the \langle111\rangle direction would require nucleation into a (111) plane followed by subsequent addition of atoms. The nucleation step allows for the possibility of twinning.

Spaepen constructs a model consisting of five to eightfold rings in the amorphous layer on top of a crystalline layer [Spa78]. Beginning with (111) layers for the model it is shown how reordering of amorphous rings back to chair-type six fold rings is necessary thus inducing recrystallization. The stresses within damaged silicon layers are examined by Seshan and EerNisse [Ses78]. Here they find the maximum resolved shear stress on inclined (111) planes is greatest for \langle100\rangle and \langle110\rangle silicon with growth along [112] directions. They presume Shockley partial dislocations (\(b=1/6[112]\)) develop and move under higher shear stress while Frank partial dislocations (\(b=1/3[111]\)) nucleate and combine with Shockley partial dislocations upon annealing to covert into a perfect loop (\(b=1/2[110]\)). For \langle111\rangle silicon it is assumed the Frank loops remain faulted since Shockley partial dislocations are less probably under smaller shear stress. Major drawbacks to this model are the prediction of regrowth in \langle100\rangle and \langle110\rangle silicon are identical and it does not account for impurity effects.

Later proposed models usually draw from previous work. The model by Drosd and Washburn [Dro82] does this by following the work of Spaepen. In their model they find it necessary for each atom transferring to the crystalline phase to complete two bonds in the hexaring. For \langle100\rangle silicon this is simple since only 1 atom is needed to complete 2
bonds due to the angle (111) planes make to the (100) plane. This occurs for each hexaring completed. For <110> silicon however, two atoms are needed to initiate the completion of the hexaring followed by addition of 1 atom at a time until the hexaring is complete. Thus the 2-atom nucleation makes it more difficult to grow in <110> directions. The <111> direction is even more difficult requiring 3 atoms to satisfy the 2-bond requirement and initiate the hexaring growth. Following the nucleation, a 2-atom requirement is then needed to complete the 2-bond requirement. In their discussion they also show how the 3-atom nucleation raises the probability of twin formation since 3 atoms may nucleate incorrectly on the surface. Narayan and co-workers took similar approaches to regrowth by considering regrowth along <110> chains of atoms to complete the hexarings [Nar82e, Nar83]. They assume that diffusion in the amorphous phase takes place and rearrangement into the crystalline material completes the hexaring making this a transfer-limited model. In their discussion they derive a more involved expression for the regrowth velocity.

\[
v = v_0 f \lambda e^{-\frac{E_a}{kT}} \left(1 - e^{-\frac{\Delta G}{kT}}\right)
\]

Here \(v\) is the regrowth velocity, \(v_0\) is the lattice vibration frequency, \(f\) is the fraction of sites available for atoms, \(\lambda\) is the atomic jump distance across the interface, \(E_a\) is the activation energy, \(k\) is Boltzmann’s constant, \(T\) is temperature in K, and \(\Delta G\) is the net free energy change. Narayan suggest \(\Delta G \approx -0.1\) eV for intrinsic silicon so the term in parentheses is close to unity for a wide range of temperatures.

Williams and Elliman also draw upon the work of Spaepen in their model. They view the process as bond-breaking and growth along [110] ledges from kink-like growth
sites [Wil83a]. Mosley and Paesler consider a dangling bond model as well. In their work they believe it is the diffusion of dangling bonds through the amorphous phase that reach the α/c interface and allow the crystalline side to capture the atom [Mos84, Pae87]. An equation similar to 2.2 is used in their analysis. Finally, Lu et al. provide a more recent review of SPE regrowth with special attention given to models by Williams and by Spaepen. They believe the dangling bonds at the α/c interface dictate regrowth phenomena [Lu91]. From their discussion they arrive at an expression for regrowth that is more involved:

\[ v = 2\sin(\theta)\nu_s n_r e^{\frac{\Delta S_f + \Delta S_m}{k}} e^{\frac{\Delta H_f + \Delta H_m}{kT}} \]  

where \( \theta \) is the angle between the interface and (111) planes, \( \nu_s \) is the speed of sound, \( n_r \) is the net distance of migration, \( \Delta S_f \) and \( \Delta H_f \) are standard entropy and enthalpy of formation of a pair of dangling bonds respectively, and \( \Delta S_m \) and \( \Delta H_m \) are entropy and enthalpy of motion of a dangling bond at the interface respectively. While the models shown attempt to explain experimental observations it is clear that there is still some doubt to the exact nature of the regrowth process.

**Impurity Considerations in SPE Models**

Doping obviously has an impact on the characteristics of SPE growth thus making models for regrowth more complicated. Looking at equation 2.2 one sees that alteration of parameters such as \( E_a \) and \( \Delta G \) leads to changes in \( v \). Lowering \( E_a \) will result in larger \( v \), on the other hand, it is pointed out by Narayan [Nar82d] that \( \Delta G \) may go to 0 if impurity concentrations are too high, pushing \( v \) to 0. To account for some of the observed phenomena several models are reviewed here.
The possibility of point defects controlling the regrowth rate is one scenario for models. Early hypotheses suggest annealing of amorphous layers creates excess vacancies that allow increased substitutional concentrations of dopants [Tok78]. Indeed it is shown that doping of silicon increases self-diffusion and thus vacancies [Fai67], so Suni et al. speculate charged vacancies are responsible for the enhancement of regrowth [Sun82b]. With increased doping, the Fermi level will shift away from midgap and in turn increase the number of charged vacancies. This model is rejected by several authors [Lu91, Par88] as they find it predicts the wrong concentration dependence.

The work of Mosley and Paesler [Mos84, Pae87] as well as Licoppe and Nissim [Lic86] illustrate the possibility of an electric field being established across the α/c interface due to band bending since the Fermi level of amorphous material is pinned at midgap. Mosley and Paesler interpret charged dangling bonds at the interface responsible for enhanced regrowth due to the electric field attracting charged species and reducing the barrier for bond rearrangement [Mos84, Pae87]. Licoppe and Nissim also assume a charge defect is influence by the generated electric field and enhanced regrowth occurs [Lic86].

Studies by Park et al. [Par88] and Jeon et al. [Jeo89, Jeo90] describe similar effect of charged defects enhancing regrowth rate. They do not go into great detail of how or by what defects the regrowth occurs but they do correlate the growth rate dependence to impurity concentration. It is also suggested the charged defect densities are controlled by the band structure of the amorphous phase. From their observations they find a dependence of $v$ by:
\[ \frac{v}{v_i} = 1 + \frac{N}{N_i} \]  

where \( v_i \) is the intrinsic silicon regrowth velocity, \( N \) is the impurity concentration, and \( N_i \) is the intrinsic carrier concentration. This dependence is fit to data for B, P, and As with good agreement as long as impurity concentration is not too high.

The approach taken by Williams and Elliman [Wil83a] and Lu et al. [Lu91] follow a model based on charged kink-like growth sites enhancing regrowth. As mentioned earlier, Williams and Elliman base their model on (111) terraces growing along [110] ledges from the propagation of kink-like sites. If an impurity is pinned at such a site it may form charged kink-like sites, which will have a kink related defect level in the band gap. Williams also presents the case of band bending at an \( \alpha/c \) interface and develops an expression for regrowth velocity based on concentrations of impurities and defect levels. The equation for donors is:

\[
\frac{v}{v_i} = \left( \frac{N}{N_c} \right) e^{\frac{E_g - E_{KD}}{kT}}
\]

where \( N_c \) is the density of states in the conduction band, \( E_g \) is the band gap energy, \( E_{KD} \) is the kink-related defect level of donors, \( g \) is the degeneracy, and \( E_{Fi} \) is the Fermi level of intrinsic silicon. A similar expression holds for acceptors and the above expression reduces to:

\[
\ln \left( \frac{v}{v_i} \right) = \ln N_D + C
\]

where \( C \) is a constant. Lu et al. follow a similar approach with fewer assumptions than Williams in developing the model [Lu91].
The various models reviewed each take their own approach to justify regrowth enhancements. For the most part it is agreed the enhancement is related to charged defects. The main question is still what the defect is and how it is produced in silicon. For now, it seems more work is necessary to fully understand the mechanism at work.

**Ion Beam Induced Crystallization**

A brief discussion of ion beam assisted growth is given here for background information on the subject. Thermal SPE regrowth is examined in this dissertation, but comparisons of results and models of ion beam techniques may be useful. Results show ion beam induced crystallization allows regrowth of amorphous layers on crystalline substrates without a dependence on substrate orientation or doping levels [Ell87a, Ell87b, Wil85]. The regrowth commences at temperatures in the range of 200-400 °C with an activation energy of 0.24 eV. It is also possible to adjust process conditions such that the amorphous layer will grow from atoms in the crystalline phase. The authors of this work suppose that the ion irradiation creates nucleation sites at the interface for regrowth. It is guessed that a vacancy or other defect is nucleated and causes atom migration or bond rearrangement. They compare this to a thermal process which has a higher activation energy but to thermally activate a defect nucleation site may take ~2.4 eV. The remaining 0.24 eV would be the barrier for the amorphous to crystalline transformation to complete. This approach suggests much of the energy added to a system is used to generate a defect so regrowth commences. It is an interesting view of the SPE process but further investigation would be warranted to understand the anomalies detected.
Table 2.1. Quoted impurity concentrations and covalent radii from various authors and references therein.

<table>
<thead>
<tr>
<th>Species</th>
<th>RBS Concentration (cm$^{-3}$)</th>
<th>Electrical Concentration (cm$^{-3}$)</th>
<th>Equilibrium Concentration (cm$^{-3}$)</th>
<th>Covalent radius (Å)</th>
<th>Reference</th>
</tr>
</thead>
<tbody>
<tr>
<td>B</td>
<td>$&gt;$10$^{20}$</td>
<td>1x10$^{20}$</td>
<td></td>
<td>0.88</td>
<td>Cse77</td>
</tr>
<tr>
<td>B</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>Tsa79a</td>
</tr>
<tr>
<td>P</td>
<td>$&gt;$10$^{20}$</td>
<td></td>
<td></td>
<td></td>
<td>Cse77</td>
</tr>
<tr>
<td>As</td>
<td>9x10$^{21}$</td>
<td>1.5x10$^{21}$</td>
<td>1.18</td>
<td></td>
<td>Wil82b</td>
</tr>
<tr>
<td>As</td>
<td>6.0x10$^{21}$</td>
<td></td>
<td></td>
<td></td>
<td>Nar83</td>
</tr>
<tr>
<td>As</td>
<td></td>
<td>5x10$^{20}$</td>
<td></td>
<td></td>
<td>Lie81</td>
</tr>
<tr>
<td>In</td>
<td>5x10$^{19}$</td>
<td>8x10$^{17}$</td>
<td>1.44-1.54</td>
<td></td>
<td>Wil82b</td>
</tr>
<tr>
<td>In</td>
<td>5.5x10$^{19}$</td>
<td></td>
<td></td>
<td></td>
<td>Nar83</td>
</tr>
<tr>
<td>Ga</td>
<td>2.5x10$^{20}$</td>
<td>4.5x10$^{20}$</td>
<td>1.26</td>
<td></td>
<td>Nar83</td>
</tr>
<tr>
<td>Sb</td>
<td>1.3x10$^{21}$</td>
<td>7x10$^{19}$</td>
<td>1.36</td>
<td></td>
<td>Wil82b</td>
</tr>
<tr>
<td>Sb</td>
<td>1.3-2.5x10$^{21}$</td>
<td></td>
<td></td>
<td></td>
<td>Nar83</td>
</tr>
<tr>
<td>Pb</td>
<td>8x10$^{19}$</td>
<td>8x10$^{17}$</td>
<td>1.54</td>
<td></td>
<td>Wil82b</td>
</tr>
<tr>
<td>Bi</td>
<td>9x10$^{19}$</td>
<td>8x10$^{17}$</td>
<td>1.46-1.56</td>
<td></td>
<td>Wil82b</td>
</tr>
<tr>
<td>Bi</td>
<td>4.5x10$^{20}$</td>
<td></td>
<td></td>
<td></td>
<td>Nar83</td>
</tr>
<tr>
<td>Tl</td>
<td>4x10$^{19}$</td>
<td>$&lt;$5x10$^{17}$</td>
<td>1.57</td>
<td></td>
<td>Wil82b</td>
</tr>
</tbody>
</table>
Figure 2.1. Correlation of maximum observed substitutional concentration to size difference of covalent radii for various impurities.
Figure 2.2. Energy diagram for transformation from amorphous to crystalline phase. Activation energy, $E_a$, is barrier to overcome and $\Delta G$ is net free energy change.
CHAPTER 3
MEASURING BORON ACTIVATION DURING SPE REGROWTH

The preliminary work in this chapter explores boron activation due to the solid phase epitaxial (SPE) regrowth process. Reviews of SPE regrowth experimental variables and characterization techniques yield insight into necessary steps for examining regrown silicon materials. The limitations of the various characterization schemes are identified here to help with the interpretation of results in this and later chapters. Findings show evidence for boron activation during SPE regrowth and discuss evaluation of data from experiment. The shortcomings of initial results are included with trends observed.

Experimental Solid Phase Epitaxial Regrowth

The SPE process is shown to incorporate and activate dopants to high levels as discussed in chapter 2. To review, the surface of crystalline silicon is amorphized with a heavy ion, typically silicon or germanium, followed by implantation of other impurities that are to be electrically activated. The silicon is then heated to temperatures of \( \geq 450 ^\circ C \) allowing the amorphous layer to recrystallize using the underlying crystalline substrate as a seed. While recrystallization takes place, impurity atoms assimilate onto substitutional lattice sites and become electrically active centers. Figure 3.1 shows a schematic of the steps involved in SPE regrowth.

The work in this dissertation focuses on examining electrical and structural characteristics of regrown amorphous layers implanted with boron. To limit diffusion of boron, regrowth temperatures remained in the 500-650 \(^\circ C\) range and times vary due to both the Arrhenius relationship of regrowth velocity and thickness of amorphous layer.
Predicting a time to complete regrowth is also complicated owing to the changing concentrations of an implant profile. Rough estimates of regrowth time are made based on regrowth rates previously measured, and ranges of time are chosen to examine regrown layers just before and after complete regrowth. The work herein uses furnace and rapid thermal annealing (RTA) techniques to regrow amorphous layers. Experiments are set up to survey ultra shallow junction (USJ) materials as possible alternatives for future microelectronic devices and deeper junction materials are used to determine maximum activation levels. Work also includes leakage current studies with relation to end-of-range (EOR) damage by quantifying defects generated from SPE regrowth. To begin however, the initial investigation verifies the ability to activate boron during SPE regrowth.

**Characterizing SPE Regrown Material**

Various electrical and structural techniques are used to characterize properties of regrown materials. A quick overview of each technique is offered to familiarize the reader with capabilities and limitations. Structural analysis consists of variable angle spectroscopic ellipsometry (VASE) and transmission electron microscopy (TEM) while electrical results come from four-point probe and Hall effect. Secondary ion mass spectrometry (SIMS) is used as well to obtain further chemical information about the material. Reference books on each technique are available for more detailed descriptions.

**VASE**

Variable angle spectroscopic ellipsometry is often used to determine optical properties of materials. For this work, layers of amorphous silicon and silicon dioxide (SiO₂) on top of crystalline silicon are of primary interest, specifically the thickness of each layer. The details of calculating the thickness of defined layers are described in appendix A. For the
most part, oxide layers are typically fixed at 20 Å (a common native oxide thickness) so measurements of amorphous layer thickness trends are more consistently found. While VASE lends itself to being a very precise measurement technique, the accuracy may be in question at times. To increase accuracy, measurements are taken at 3 angles (65, 70, and 75°) so models are fit to larger amounts of data. Scanning over a larger range of wavelengths of light also increases accuracy of measurements. Obstacles in exactness arise from incomplete amorphization near the amorphous/crystalline (α/c) interface leading to rough interfaces, which further complicate light reflection. Impurity additions change dielectric constants used in modeling data thus affecting accuracy in measurements as well. Overall the VASE system normally yields an accuracy of 10-20% of actual thickness determined from TEM.

**TEM**

Defect and layer information is assessed using TEM. Use of plan-view TEM (PTEM) provides knowledge on defect density and size. Three millimeter diameter cores cut from silicon pieces are mechanically thinned and etched with 3:1 HF:HNO$_3$ solution to create thin regions for imaging in TEM. Imaging is done using $g_{220}$ weak beam dark field (WBDF) and $g_{3g}$ diffracting conditions to optimize the contrast of defects observed. Defects are then quantified using techniques described elsewhere [Bha97]. Cross-sectional TEM (XTEM) samples are prepared by cutting strips of silicon then gluing them together. Cores 3 mm in diameter are cut and thinned mechanically, then ion-milled to sufficiently thin layers for imaging. Images are taken from the [110] zone using bright field (BF) to measure thickness of layers and depth of EOR damage. In some occasions it is necessary to use high resolution XTEM (HR-XTEM) to determine distance
of very thin layers. Although sample preparation is a tedious process, TEM is a well-established technique for characterizing materials.

**SIMS**

Knowledge of chemical information for implanted impurities comes from SIMS results. Measurements consist of sputtering a crater in the surface of the sample while counting the number of impurity atoms sputtered with time. Then measuring the crater depth yields information of impurity concentration versus depth into the substrate (assuming constant sputter rate) if the counts measured are compared to counts determined from known impurity concentrations. Errors from uneven sputtering of the crater lead to incorrect concentration profiling. Minimum detected boron impurity concentration limits are regularly \(10^{16}-10^{17}\) cm\(^{-3}\) using an O\(_2\) source beam. While impurity concentration profiles are measured, no information about substitutional or interstitial positioning is distinguished.

**Four-Point Probe**

Four-point probe is a simple technique for measuring electrical characteristics to calculate sheet resistance or resistivity. The system relies on a set of four in-line metal tips coming in contact with the samples. A set current is forced through the two outside probes while voltage is measured using the two inside probes. Then sheet resistance \(R_s\) is calculated from:

\[
R_s = F \frac{V}{I}
\]

where \(V\) is voltage measured, \(I\) is the set current, and \(F\) is a factor based on geometry. For samples having large diameter or side length compared to the overall probe width \(F=\pi/\ln(2)\), but for smaller samples the factor decreases [Per77]. Samples are measured
with forward and reverse current to get an average for greater accuracy. For well-behaved samples, measurements taken at separate incidences should be within 10% of each other. Using four-point probe gives a quick and easy measurement of basic electrical characteristics for implanted layers.

**Hall Effect**

Many of the electrical characteristics of a material come about from Hall measurements. Sheet resistance, active dose/concentration, and mobility are all calculated from Hall measurements, but only if samples are prepared correctly. Most important is contacting the sample by choosing the appropriate metal for ohmic contacts and placing contacts in correct positions. A mathematical treatment by van der Pauw explains why contacts to test specimens need to be relatively small and on the perimeter to determine electrical properties from arbitrary shaped samples [van58]. The standard shape of the specimen should be symmetrical such as a square or clover-leaf [AST89], however it is generally easier to cut squares than manufacture clover-leafs. For square samples, edge lengths of $\geq 15$ mm provide accurate measurements [AST89, Sch98] but smaller edge lengths lead to reliable results if contacts are small and placed at the corners. Figure 3.2 to 3.4 illustrate the effect edge length has on active dose, mobility, and sheet resistance for silicon samples preamorphized with $30$ keV, $1 \times 10^{15}$ cm$^{-2}$ Si$^+$ and implanted with $5$ keV, $2 \times 10^{15}$ cm$^{-2}$ B$^+$ followed by a $1000$ °C, $60$ min anneal in a nitrogen purged quartz tube furnace. Placing indium (In) metal on to the corners and pressing until intimate contact to silicon is made creates ohmic contacts. Within a 10% uncertainty of the measurements, active dose, mobility, and sheet resistance are about constant over the range of edge sizes. An aluminum/silicon alloy (96.5 wt% Al/3.5 wt% Si) is also used for ohmic contacting but needs to be selectively evaporated onto samples and may need
to be annealed at moderate conditions (e.g. 450 °C for 30 min) for ohmic formation. Aluminum contacted samples having edge length ≥14 mm produce consistent measurement of electrical characteristics, but these samples have evaporated Al/Si away from the corners, thus producing untrue van der Pauw geometry, so larger samples reduce errors from edge effects. The Al/Si alloy is used rather than pure Al because it reduces the spiking of Al into the silicon substrate [Cam96], which needs to be minimized so contact to both the active p-type layer and n-type substrate is avoided. Contacting to both the layers leads to errors in measurements. Several other metals are suitable to contact p-type silicon [Li93] but In and Al/Si are common so are used for samples examined in this work.

Although proper sample preparation leads to consistent results, interpretation of data may invite incorrect assumptions of properties. Simple Hall measurements do not consider effects of light-hole, heavy-hole, and spin-orbit split bands together, therefore correction factors need to be added into data [Li79, Lin81]. Li derives an expression for the Hall coefficient, $R_H$, that includes the Hall factor, $r$ [Li93].

$$R_H = \frac{r}{qp}$$

This expression is for p-type doping but a similar expression is obtained for n-type. Here $q$ is the charge of the hole (or electron) and $p$ is the active concentration of holes (or electrons). Expressions to calculate properties such as activation and mobility are shown in appendix B but do not include $r$ for simplification. To obtain true values of dose activation one needs to multiply by the Hall factor while conductivity mobility is determined by dividing Hall mobility by the Hall factor. The Hall factor depends on several variables such as active concentration and magnetic field strength so difficulty in
obtaining an accurate value persists. Experimental results show the Hall factor varies from \( \sim 0.8 \) for boron concentrations around \( 10^{18} \text{ cm}^{-3} \) to \( \sim 0.7 \) for concentrations \( >10^{20} \text{ cm}^{-3} \) \cite{Li79, Lin81}. High magnetic fields force the Hall factor to unity thus allowing true activity and conductivity mobility to be determined. Large magnetic fields are not attainable in the work contained within so a Hall factor of 0.7 is assumed as a conservative estimate since active boron concentrations \( >10^{20} \text{ cm}^{-3} \) are realized.

**Activating Boron During SPE Regrowth**

To examine activation during regrowth requires electrical contacts as deep as the initial \( \alpha/c \) interface so any distance of regrown material will be included in electrical measurements (it is assumed the amorphous material is electrically inactive). This is accomplished by lithographically defining contacts on silicon samples. N-type silicon pieces cut 7 mm x 7 mm are covered with photoresist except near the corners. Implantation of 30 keV, \( 1 \times 10^{15} \text{ cm}^{-2} \) \( \text{B}^+ \) into the samples leaves boron in the corners only, which is then activated using a 900 °C, 30 min anneal. Photoresist is then stripped and placed onto the corners of the sample as preparation to further implanting.

Preamorphizing the samples with 30 & 90 keV, \( 1 \times 10^{15} \text{ cm}^{-2} \) \( \text{Si}^+ \) followed by a 30 keV, \( 1 \times 10^{15} \text{ cm}^{-2} \) \( \text{B}^+ \) leaves the samples ready to be regrown with crystalline silicon contacts at the 4 corners of the sample. Then Al/Si alloy is evaporated onto the corners of samples using an electron-beam evaporator and a shadow mask. Following evaporation, a 450 °C, 30 min anneal allows ohmic contacting to the crystalline silicon corners. A schematic of this process is given in figure 3.5. Now the sample is ready for regrowth studies.

Since electrical contacts now extend down to the original \( \alpha/c \) interface this gives a convenient method to measure Hall characteristics as the amorphous layer regrows. Annealing of a sample is performed at 500 °C in 6-minute intervals and larger times after
the layer is nearly regrown. Amorphous layer thickness is measured using J A Woollam M-88 VASE system while electrically activated dose is monitored using an MMR Hall effect system at 3000 G. The Al/Si contacts are ~1 mm from the edges of the sample and are ~1.5 mm in diameter, thus there is not a true van der Pauw geometry, generating error in Hall measurements. Figure 3.6 shows results for percentage of adjusted active boron dose versus the anneal time along with the amorphous layer thickness versus anneal time. The Hall results measure a maximum dose ~7 times higher than the implant dose (due to the incorrect contact geometry) when there is ~370 Å of amorphous layer left, therefore Hall results are scaled by this maximum for qualitative comparison. Also included in figure 3.6 is the percentage of boron in the regrown crystalline silicon versus time measured by integrating the SIMS profile. The SIMS profile is given in figure 3.7 and shows the depth of the amorphous layer for each time interval denoted by the vertical lines. The activation curve in figure 3.6 agrees fairly well with the known percentage in the regrown layer.

The activation trend in figure 3.6 reaches maximum activation before the amorphous layer is completely regrown. Further annealing shows a slight decline in active levels suggesting slight deactivation but longer times show no further lowering in activation. At longer anneal times the activation levels fluctuate some and the error bars indicate measured active levels are within error of the maximum measured active dose. Error bars represent a 10% uncertainty as typically seen in Hall measurements, however error may be larger in this sample due to incorrect van der Pauw arrangement. Therefore it is reasonable to assume little deactivation occurs and further testing using true van der Pauw geometry would lead to accurate results.
Summary

A review of the SPE regrowth process and its effect on activating boron is presented here. Details of experimental considerations to study regrowth and activation are shown along with brief descriptions of common characterization techniques employed. Gathering and analyzing data demonstrates difficulties and limitations to each of the measurements used in studies. Conclusive evidence indicates boron does activate during the SPE regrowth process and the lithographically defined contacting process is a very convenient method for measurements. Control of sample and contact size is shown to be an important factor when analyzing samples using Hall systems and may lead to error in data if proper steps are not taken.
Figure 3.1. Schematic of experimental set-up for experiments involving solid phase epitaxial regrowth showing a) the amorphization step and b) dopant introduction followed by c) SPE regrowth at low temperatures.
Figure 3.2. Active dose as a function of square sample size. Samples are implanted with 5 keV, $2 \times 10^{15}$ cm$^{-2}$ B$^+$ and annealed at 1000 °C for 60 min.
Figure 3.3. Hall mobility as a function of square sample size. Samples are implanted with 5 keV, 2x10^{15} cm^{-2} B^+ and annealed at 1000 °C for 60 min.
Figure 3.4. Sheet resistance as a function of square sample size. Samples are implanted with 5 keV, 2×10^{15} cm^{-2} B^+ and annealed at 1000 °C for 60 min.
Figure 3.5. Schematic representation of the lithographically defined contacting process where a) boron is implanted into corners of crystalline silicon and annealed followed by b) preamorphization and dopant introduction everywhere except active corners. A shadow mask is placed on samples and Al/Si is evaporated onto the crystalline corner contacts, c), and finally d) annealing of the amorphous layer recrystallizes the silicon.
Figure 3.6. SPE regrowth at 500 °C of a sample preamorphized with 30 & 90 keV, 1x10^{15} \text{ cm}^{-2} \text{ Si}^+ and implanted with 30 keV, 1x10^{15} \text{ cm}^{-2} \text{ B}^+. Boron activation is shown as the amorphous layer is regrowing.
Figure 3.7. SIMS profile for 30 keV, $1 \times 10^{15}$ cm$^{-2}$ boron with vertical lines indicating depth of amorphous/crystalline interface during annealing. Area under the SIMS curve for regrown portion of silicon is calculated at each anneal interval to determine percentage of boron in regrown layer.
CHAPTER 4
SOLID PHASE EPITAXIAL REGROWTH OF ULTRA SHALLOW JUNCTION MATERIALS

Introduction

Getting to future technology nodes on the International Technology Roadmap for Semiconductors (ITRS) [ITR01] requires developing innovative procedures to create ultra-shallow, highly active layers in silicon. According to the next couple years in the ITRS, junctions will need to reach depths of 15-31 nm while sheet resistances of less than $660 \, \Omega/\square$ are desired [ITR01]. Figure 1.2 shows the complete requirements for the ITRS. While high activation of dopants is required (achievable through conventional techniques) it also means diffusion needs to be minimized for the formation of ultra-shallow, abrupt junctions. Laser annealing has the capability of producing the desired ultra-shallow junction (USJ) material but continues to face integration problems into integrated circuit (IC) manufacturing. Solid phase epitaxial (SPE) regrowth at low temperatures affords the opportunity for generating shallow, active layers by incorporating dopants as the amorphous layer crystallizes. Using temperatures $\leq 650 \, ^{\circ} \text{C}$ but $>450 \, ^{\circ} \text{C}$ reduces diffusion of dopants, yielding better control of junction depth, while allowing recrystallization. Preamorphizing helps reduce junction depth and improve the junction abruptness as well by eliminating ion channeling if the dopant implant is contained within the amorphous layer. Thus the SPE regrowth process promises to create active layers and carrier profiles that follow relatively close to as-implanted chemical profiles.
Observations of the SPE regrowth process are largely researched in the late 1970 to early 1990s but current limitations are renewing its interest. Recent investigations into USJ materials are beginning to show it is possible to use ultra-low energy (ULE) implants or plasma doping (PLAD) with SPE for meeting future requirements in IC processing [Har98, Lin00, Lin02, Osb96]. As boron is often the limiting dopant it is the main focus of these SPE studies. Indeed it is possible to integrate a SPE process into complimentary metal oxide semiconductor (CMOS) device manufacturing since the required equipment already exists in most fabs. Several groups have already shown it is possible to fabricate SPE processed CMOS devices with acceptable characteristics making use of silicon-on-sapphire (SOS) substrates [May90, Zho98], lateral SPE on silicon-on-insulator (SOI) substrates [Hir89, Kum01, Kum02, Liu01, Liu02], and more common SPE techniques [Mit96, Miy00, Tsu99]. Although SPE manufactured CMOS devices may yield acceptable performance it is worthwhile to examine USJ materials at a more fundamental approach to understand conditions of optimum performance.

**Experimental**

General experimental set-up consists of n-type, <100> silicon wafers all implanted with the same preamorphizing condition and variable boron dose. Wafers are preamorphized with 2.5 keV, 1x10^{15} \text{ cm}^{-2} \text{ Si}^+ using a deceleration mode of 7.5 keV to 2.5 keV and beam current of 1798.5 \mu A with a standard deviation of 6.5 \mu A. Boron is implanted at 0.5 keV and doses of 1-9x10^{15} \text{ cm}^{-2} using average beam current of 1797.5 \mu A with standard deviation of 10.8 \mu A. Following the implants, pieces are cut 14 mm by 14 mm and cleaned with water and acetone.

Annealing of samples takes place in a quartz tube furnace under flowing nitrogen (N_2) gas. The quartz boat remains in the furnace until samples are to be placed in it for
annealing. Using a quick push/pull method, sample pieces are annealed between 500 and 650 °C for 0.5 to 6 minutes. Select samples are annealed at longer times when necessary to further study regrowth. Because these times are short for furnace annealing, temperature profile measurements are gathered to determine actual temperatures. A thermocouple (held on the furnace wall before the boat arrives in the furnace) is placed on the quartz boat and temperature is recorded versus time. Figure 4.1 illustrates temperatures set points of 500-650 °C in 50 °C intervals. Initial decrease in temperature is the time it takes the thermocouple to reach the temperature of the boat but shows the thermocouple has reached that of the boat after ~0.5 minute, which is ~25-30 °C below the furnace temperature. Reaching the desired temperature takes just over 2 minutes suggesting only longer times allow samples to reach the desired temperature.

To determine the success of amorphous layer regrowth requires the use of variable angle spectroscopic ellipsometry (VASE) and four-point probe. The VASE is a J A Woollam M-88 system with measurements taken at 65-75°. High resolution cross-sectional transmission electron microscopy (HR-XTEM) on a JEOL 2010F system allows comparison of VASE results to true amorphous layer thickness. The Jandel Engineering Ltd. four-point probe system uses 100 µm diameter in-line tungsten-carbide tips spaced ~1 mm apart and 60+ g force. Current is fixed at 100 µA and each sample is measured 5 times (each measurement uses a forward and reverse current to average voltage) with sample rotation between measurements. The multiple measurements give a standard deviation or uncertainty of the measurement.

Once initial VASE and four-point probe measurements are taken, samples are prepared for Hall measurements. Preparing samples for Hall measurements requires
metal contact formation at or near the corners of samples. Samples are cleaned in acetone, methanol, trichloroethylene (TCE), and water before removing the native oxide with 10:1 HF:NH$_4$F buffered oxide etch (BOE). Aluminum/silicon alloy (96.5 wt%Al/3.5 wt% Si) is then evaporated about 1 mm from the edges in ~1 mm diameter dots using electron-beam evaporation. All samples are then annealed again at 450 °C for 30 minutes in the quartz tube furnace to help ohmic contact formation. Hall measurements are taken on the MMR Technologies system (MPS-50, K-20, and H-50) with a field of 3000 G and current maximized by the system. The Hall system determines maximum current as 85% of the current across the contact pair that shows the minimum current needed to sustain a 1.5 V potential. As discussed in chapter 3 the Hall factor is estimated to be 0.7 since active concentrations exceed $10^{20}$ cm$^{-3}$ holes [Lin81].

**Results**

Recrystallized layers are first examined by VASE to examine the extent of regrowth for all samples. Figures 4.2 and 4.3 illustrate the amorphous layer thickness results from samples annealed at 500-650 °C for 0.5 to 6 minutes with some samples at 500 °C annealed at longer times (up to 40 minutes). The regrowth occurs very quickly for all samples and is essentially complete after about 2 minutes or about 4 minutes for annealing at 500 °C. It is noted, however, that samples implanted with a dose of $5 \times 10^{15}$ cm$^{-2}$ or above do not allow full recrystallization of the amorphous layer. A residual amorphous layer between ~10-30 Å thick always remains for these higher doses. The Stopping and Range of Ions in Matter (SRIM) simulation at 0.5 keV is run and doses are corrected for sputter loss [Aga00] so estimates of the concentration profiles establish approximate boron concentrations. A $5 \times 10^{15}$ cm$^{-2}$ boron dose creates a peak boron concentration of ~$5 \times 10^{21}$ cm$^{-3}$ which is consistent with previous work showing boron
concentrations >4x10^{21} \text{ cm}^{-3} \text{ significantly reduce and halt regrowth} \ [\text{Ker84}]. \text{ Annealing for longer times at 500 } ^\circ \text{C and 600 } ^\circ \text{C (not shown) has little effect on the thickness of the amorphous layer. High resolution XTEM confirmed complete regrowth at 1x10^{15} \text{ cm}^{-2} \text{ boron dose and realization of a 24 } \text{Å} \text{ amorphous layer in 9x10^{15} cm}^{-2} \text{ boron for samples annealed at 500 } ^\circ \text{C for 6 minutes. Images from the microscope are given in figure 4.4. The VASE calculated an amorphous layer 13 Å thick for a fixed oxide of 20 Å so some error persists in VASE measurements when probing very thin layers. Samples at the high dose are also annealed up to 1000 } ^\circ \text{C and still indicate the presence of an amorphous layer from VASE findings.}

Starting with sheet resistance measurements, four-point probe is used before further processing for Hall measurements. Four-point probe data is limited to samples annealed at 600 and 650 °C as measurements at lower anneal temperatures yielded unstable results, presumably from non-ohmic contact to silicon. Figure 4.5 indicates the trends in sheet resistance, $R_s$, as a function of anneal time for the various implant doses. As expected, sheet resistance generally decreases with anneal time, which is consistent with regrowth of the amorphous layer. However, the smallest dose (1x10^{15} \text{ cm}^{-2}) \text{ follows the opposite trend by increasing } R_s \text{ as anneal time becomes larger. The } R_s \text{ data from Hall measurements acknowledges similar trends to those of four-point probe as seen in figures 4.6 and 4.7. The error bars in Hall measurements come from the standard deviation seen in the measurement of 10 samples annealed at 600 °C for 4 minutes. Hall measures electron carriers for samples with dose of 1x10^{15} \text{ cm}^{-2} \text{ boron annealed at 500 and 550 °C, for all times, as well as 600 and 650 °C for 0.5 minute, thus results are omitted since they are not consistent with hole measurements. The reason for electron carrier outcome is}
not entirely clear but it is suspected either Al has spiked into the substrate [Cam96] or leakage currents are large causing error in measurements. This is discussed further in chapter 6. Data is a little more scattered for samples annealed at 500 °C, but higher temperatures show $R_s$ decreasing and leveling off at higher anneal times. Taking the data for $R_s$ of samples annealed at 600 and 650 °C for 4 min and plotting versus boron implant dose provides the trend seen in figure 4.8. Clearly increasing the B dose only increases the sheet resistance leading to degraded material property. Both data from the four-point probe and Hall system are plotted and are fairly consistent with one another, though the Hall data is somewhat more scattered. Theoretical predictions of $R_s$ are included in figure 4.8 too, where calculations assume 100% activation of boron in the concentration profile estimated from SRIM. A model to estimate mobility by Masetti et al. [Mas83] is used with the concentration profile to model a set of parallel sheet resistances and calculate overall $R_s$. Not only are the theoretical values much lower than actual values but also sheet resistance should decrease with increasing dose. It is obvious that much of the boron is inactive for these implants and implant conditions need to be optimized by lowering the dose to minimize sheet resistance. If the Hall sheet resistance is now plotted versus anneal temperature for fixed anneal time, the result is seen in figures 4.9 and 4.10. Certainly it is shown that higher anneal temperatures reduce the sheet resistance, however even temperatures around 600 °C may lead to slight boron diffusion and greater junction depths [Lin02]. High resolution Secondary Ion Mass Spectrometry (SIMS) is not available at the time of this work so actual chemical junction depth (usually set as $1 \times 10^{18}$ cm$^{-3}$) is undetermined. From the SRIM simulation it is believed to be between 100 and 140 Å but this does not take into account energy contamination, so junctions may be
closer to 200 Å. The isochronal data in figures 4.9 and 4.10 are limited to anneals at 1-4 minutes where most regrowth has occurred.

Hall measurements verify the information gained from sheet resistance data. Figure 4.8 demonstrates how increasing boron dose raises $R_s$ concluding the excess boron does not activate upon regrowth. Active dose typically measures in the 1-1.5x10^{14} \text{ cm}^{-2} range with active levels reaching as high as ~2.2x10^{14} \text{ cm}^{-2}. Since the dose of holes is essentially constant, with only minor fluctuation, these high implant doses are not activating the majority of dopant. This is seen in figures 4.11 and 4.12, which plot the percent activated boron as a function of implant dose. At the lowest dose ~15% of the boron is active but quickly drops to $\leq 5\%$ as boron dose is increased. As mentioned earlier, the actual retained dose is typically much smaller than the implanted dose due to sputter loss at low energy implantation [Aga00] so percent activation is based on estimates of retained dose. Increasing anneal temperature decreases sheet resistance, as seen in figures 4.9 and 4.10, therefore either mobility or active concentration will rise. If active levels are essentially constant then it follows mobility must be slightly increasing with anneal temperatures. Hole mobility versus anneal temperature given in figures 4.13 and 4.14 for fixed anneal times confirmed the slight mobility increase with anneal temperature.

**Discussion**

Although high concentrations of boron permit large hole concentrations to be achieved using SPE regrowth, continued increase in boron levels is detrimental to electrical properties. The lowest boron dose in the above experiment yields the best electrical characteristics while also having the smallest junction depth. Since maximum active doses reach ~2x10^{14} \text{ cm}^{-2} one could assume optimized $R_s$ will exist for doses below
1x10^{15} \text{cm}^{-2} \text{ and above } 2x10^{14} \text{cm}^{-2}. \text{ This work contradicts the findings of Mizushima et al. who argue the existence of icosahedral B_{12} complexes at high boron concentrations are responsible for a very low resistive phase in silicon [Miz93, Miz94]. The claim is the icosahedra act as double acceptors and form at room temperature during high dose boron implantation. However, the boron-silicon phase diagram [Ole84] reports a SiB_{3} phase would form before complexes of 12 boron atoms. A SRIM simulation of their implant conditions estimates peak boron concentrations between 3x10^{21} \text{ and } 1x10^{22} \text{cm}^{-3}. The SRIM simulation for 0.5 keV boron (with sputter loss correction) indicates peak boron concentrations in the range of 1-7x10^{21} \text{ cm}^{-3}, which is close to that of Mizushima et al. The formation of a low resistive phase in the above experiment is not observed and increasing boron concentrations only has a negative effect as expected.

Estimates of the maximum hole concentrations are extracted from Hall data since direct measurement of such shallow layers using techniques such as spreading resistance profiling (SRP) would be difficult. The concentration profiles obtained from the 0.5 keV SRIM simulation permit comparison to measured active boron dose measured from Hall. The concentration profiles are modified such that boron is assumed active up to a maximum concentration and any boron above this concentration is assumed inactive. The dose integrated from this modified profile matches the dose measured by the Hall system so the approximate maximum achievable hole concentration is determined. A macro program written in Visual Basic is given in appendix C and is used to modify concentration profiles until the integrated dose matches the Hall dose input by the user. Once the modified profile is found a theoretical Hall coefficient is calculated from:
Details of the equation derivation are in appendix B. Here \( q \) is the charge of a hole \((1.6 \times 10^{-19} \text{ C})\), \( t \) is the total active layer thickness, \( n_p(z) \) is the hole concentration as a function of depth, \( z \), and \( \mu_p(z) \) is the mobility at \( z \) approximated from the mobility model of Masetti et al. [Mas83]. The Hall coefficient is divided by the thickness \((t)\) of the active layer to acquire the sheet Hall coefficient, which reduces a variable that would need to be established. The calculated sheet Hall coefficient is then compared to the sheet Hall coefficient measured by the Hall system. For all samples measuring holes in this experiment the difference between the calculated and measured sheet Hall coefficients is \( \leq 0.6\% \). Such close agreement between calculated and measured values suggests the SRIM concentration profiles are reasonable approximations for the samples. The initial amorphous layer is around \( \sim 100-120 \text{ Å} \) so any boron implanted deeper than that would be assumed inactive for such low annealing temperatures. As the amorphous layer reorders the boron is incorporated and activates creating a p-type layer \( \sim 100-120 \text{ Å} \) thick which would be near the same depth predicted by the SRIM profile. This makes the true active concentration profile similar to that of the SRIM explaining the good agreement between theoretical and measured Hall coefficients. From a chemical standpoint, the junctions would be deeper for higher implant dose but not if the tail is inactive. This helps explain how small increases in sheet resistance with increasing dose are due to the excess boron added to the amorphous layer. This extra boron will never become active but acts as a scattering site. Plotting the hole concentrations realized for anneals at 1-4 minutes (shown in figures 4.15 and 4.16) indicates maximum active boron concentration up to
2.2x10^{20} \text{ cm}^{-3}. Typical hole concentrations lie between \sim 1-1.6x10^{20} \text{ cm}^{-3} and do not appear to be significantly affected by implant dose or anneal temperature. However, it may be possible to attain higher hole concentrations if lower boron doses are implanted so any effects from excess boron are eliminated. Referring back to figure 1.3, these concentrations in the 10^{20} \text{ cm}^{-3} range are sufficient to meet future IC processing needs. Continued optimization of dose and $R_s$ is expected to meet direct requirements stated in the ITRS.

**Summary**

The formation of USJ material with low sheet resistance is needed for continued scale-down of MOSFET devices. Several versions of the SPE growth process have already revealed success in creating working CMOS devices. Understanding the formation of USJ material utilizing SPE regrowth is necessary to properly design future material and anticipate limitations. Adding too much boron reduces and halts the regrowth process and degrades electrical properties of regrown layers. Increasing the SPE anneal temperature reduces the sheet resistance but care must be taken to determine if boron is diffusing and piling up at the end-of-range damage. The advantage of increasing anneal temperature is revealed as slightly increased mobility while active boron levels remain about constant. Hole concentrations of \sim 1-2x10^{20} \text{ cm}^{-3} for this USJ material promise to be sufficient for future generation devices. More work on optimizing electrical characteristics is mandatory to better understand the effect SPE regrowth poses on ULE implanted silicon.
Figure 4.1. Temperature profiles for a quartz boat in a quartz tube furnace as a function of time. Equilibrium temperature is not reached until after ~2 minutes.
Figure 4.2. Regrowth of amorphous layers from 2.5 keV, $1 \times 10^{15}$ cm$^{-2}$ Si$^+$ implants with 0.5 keV B$^+$ for anneals of a) 500 °C and b) 550 °C measured from VASE. Doses above $3 \times 10^{15}$ cm$^{-2}$ do not completely regrow.
Figure 4.3. Regrowth of amorphous layers from 2.5 keV, 1x10^{15} \text{ cm}^{-2} \text{ Si}^+ \text{ implants with} 0.5 \text{ keV B}^+ \text{ for anneals of a) 600 °C and b) 650 °C measured from VASE. Doses above} 3\times10^{15} \text{ cm}^{-2} \text{ do not completely regrow.
Figure 4.4. High-resolution cross-sectional TEM images of samples preamorphized with 2.5 keV, 1x10^{15} \text{cm}^{-2} \text{Si}^+ and implanted with a) 1x10^{15} \text{cm}^{-2} \text{B}^+ and b) 9x10^{15} \text{cm}^{-2} \text{B}^+.

Samples are annealed at 500 \text{C} for 4 min. The low boron dose sample shows full regrowth of the amorphous layer while the high boron dose halts regrowth with 24 Å left.
Figure 4.5. Sheet resistance measurements from four-point probe of 2.5 keV, $1 \times 10^{15}$ cm$^{-2}$ Si$^+$ and 0.5 keV B$^+$ implants following SPE regrowth and before Hall measurements for a) 600 °C and b) 650 °C anneals.
Figure 4.6. Sheet resistance measured from Hall system for samples implanted with 2.5 keV, $1 \times 10^{15}$ cm$^{-2}$ Si$^+$ and 0.5 keV B$^+$ and annealed at a) 500 °C and b) 550 °C.
Figure 4.7. Sheet resistance measured from Hall system for samples implanted with 2.5 keV, $1 \times 10^{15}$ cm$^{-2}$ Si$^+$ and 0.5 keV B$^+$ and annealed at a) 600 °C and b) 650 °C.
Figure 4.8. Sheet resistance versus boron implant dose for 600 and 650 °C anneals at 4 minutes for preamorphization of 2.5 keV, $1 \times 10^{15}$ cm$^{-2}$ Si$^+$ and 0.5 keV B$^+$ implants illustrating higher doses lead to poorer electrical properties.
Figure 4.9. Sheet resistance as a function of anneal temperature for implants of 2.5 keV, $1 \times 10^{15}$ cm$^{-2}$ Si$^+$ followed by 0.5 keV B$^+$ at a) 1 minute and b) 2 minute anneals showing decreased sheet resistance for higher anneal temperatures.
Figure 4.10. Sheet resistance as a function of anneal temperature for implants of 2.5 keV, 1x10^{15} \text{ cm}^{-2} \text{ Si}^+ followed by 0.5 keV B^+ at 4 minute anneals showing decreased sheet resistance for higher anneal temperatures.
Figure 4.11. Estimated percent activation of SPE regrown materials with 2.5 keV, $1 \times 10^{15}$ cm$^{-2}$ Si$^+$ and 0.5 keV B$^+$ for anneals at a) 1 minute and b) 2 minutes.
Figure 4.12. Estimated percent activation of SPE regrown materials with 2.5 keV, $1 \times 10^{15} \text{cm}^{-2}$ Si$^+$ and 0.5 keV B$^+$ for anneals at 4 minutes.
Figure 4.13. Hole Mobility as a function of anneal temperature for anneals at a) 1 minute and b) 2 minutes. Samples are preamorphized with 2.5 keV, $1 \times 10^{15}$ cm$^{-2}$ Si$^+$ and implanted with 0.5 keV B$^+$. 
Figure 4.14. Hole Mobility as a function of anneal temperature for 4 minute anneals. Samples are preamorphized with 2.5 keV, $1 \times 10^{15}$ cm$^{-2}$ Si$^+$ and implanted with 0.5 keV B$^+$. 
Figure 4.15. Maximum active boron concentrations following SPE regrowth for anneals at a) 1 minute and b) 2 minutes. Preamorphization of 2.5 keV, 1x10^{15} cm^{-2} Si+ followed by 0.5 keV B+ implants.
Figure 4.16. Maximum active boron concentrations following SPE regrowth for anneals at 4 minutes. Preamorphization of 2.5 keV, $1 \times 10^{15}$ cm$^{-2}$ Si$^+$ followed by 0.5 keV B$^+$ implants.
CHAPTER 5
IMPACT OF SOLID PHASE EPITAXIAL REGROWTH ON JUNCTION LEAKAGE

Introduction

Much of the current work involving solid phase epitaxy (SPE) focuses on fabricating shallow junctions with low sheet resistance. However, a suitable active layer must also exhibit good junction leakage characteristics for use within an integrated circuit (IC) process. Thus it is necessary to vary implant conditions to examine the effect on leakage characteristics. It is known for some time that defects in the space-charge region contribute to leakage currents in bipolar transistors [Ash77, Bul79]. This damage severely degrades the performance of devices, thus it needs to be removed or minimized in the depletion layer. According to the International Technology Roadmap for Semiconductors (ITRS), junction leakage should only donate a small amount to the total leakage during the off-state of a metal oxide semiconductor field effect transistor (MOSFET) [ITR01]. This strict requirement demands junctions be well-behaved, which is not always easy considering the silicon lattice damage created during ion-implantation. Therefore it seems necessary to further explore the characteristics of junctions developed from SPE regrowth.

In order to examine the impact of SPE on junction leakage necessitates a general overview of factors contributing to leakage. Knowing that defects enhance leakage means detailed studies on defects are needed. Not only are the defects themselves important but also their proximity to the electrical junction plays a role. With this in mind some of the key variables are defect densities, interstitial densities, and defect size.
Supposing that leakage occurs at the strain within the crystalline lattice leads one to believe the line length of a defect is key. Silicon defects generally consist of loops and/or <311> defects but typical defects in SPE regrown materials are generally small dots in transmission electron microscope (TEM) images, so it is hard to distinguish between them. The perimeter of these dots represents the line length of defects for the purposes of the following experiment.

**Experimental**

N-type, 200 mm silicon wafers of 10-20 Ω-cm resistivity and <100> orientation are implanted with phosphorus to create a uniform concentration of ~5x10^{17} cm^{-3}. The implant sequence to generate this n-well is 500 keV at 2x10^{13} cm^{-2}, 250 keV at 1x10^{13} cm^{-2}, 100 keV at 5x10^{12} cm^{-2}, and 20 keV at 1.2x10^{12} cm^{-2} and is followed by furnace annealing at 800 °C for 30 minutes. Ion implants of 5-30 keV, 1x10^{15} cm^{-2} Si^+ preamorphize the surface of wafers to various depths, which are measured by an ellipsometer. Boron doping is then introduced using BF$_3$ in a Varian Semiconductor Equipment Associates pulsed plasma doping (P$^2$LAD) system at biases of 0.6-5 kV and doses of either 5x10^{15} or 2x10^{16} cm^{-2}. The conditions of the implants are given in table 5.1. An anneal at 560 °C for 15 minutes on a proprietary hot chuck at 1x10^{-6} torr vacuum completes the SPE regrowth of amorphous layers. Wafer processing is performed in external facilities. To illustrate the purpose of chosen implant conditions a schematic of the experimental set-up is specified in figure 5.1. The implant conditions place the end-of-range (EOR) damage a) before, b) at or near, or c) after the boron tail to decide the effect on junction leakage current.

Now that the experimental set-up is established it is apparent electrical, chemical, and structural characterization techniques are desirable for investigation. Electrical and
chemical procedures are measured in external facilities as well. Leakage measurements are performed using reverse bias of –1 V with a 3-probe measurement system on mesa diode structures. Two probes are placed on the implant layer where one supplies current and the other measures voltage while the third probe is placed on the silicon base to collect the leakage current. Further details of the experimental setup for leakage measurements are found in reference [Mad03]. The chemical profiles from a PHI quadrupole secondary ion mass spectrometry (SIMS) system using 1 keV O$_2^+$ primary ions at 60° angle of incidence with oxygen leak provides data to approximate junction depth by defining 1x10$^{18}$ cm$^{-3}$ boron concentration as the metallurgical junction point. Defect characterization on a JEOL 200CX TEM yielded information about defect nature and depth of EOR damage. Plan-view TEM (PTEM) images of samples are taken from $g_{3g}$ weak beam dark field (WBDF) and $g_{220}$ for defect quantification. Details of counting defects are well established and followed from the work of Bharatan et al. [Bha97]. Cross-sectional TEM (XTEM) samples on the <110> zone are imaged in bright field (BF) to resolve the depth of EOR damage.

Results

End-of-range damage depths found from XTEM are used to relate to amorphous layer thickness as well as junction depths. For higher energy implants it is expected the straggle is larger so excess interstitials from the amorphization implant will reside further from the amorphous/crystalline ($\alpha$/c) interface as implant energy is increased. Figure 5.2 conveys this notion by plotting the depth of EOR damage versus the amorphous layer thickness from ellipsometry. Included in the plot is a line of x=y which has a slope of 1. The line fit for the data indicates a slope of ~1.3 clearly demonstrating the effect higher implant energies have on pushing EOR damage further from the $\alpha$/c interface. Next, the
distance of the junction, $x_j$, relative to the EOR damage depth, $x_{EOR}$, is compared to the leakage current observed in samples. This is represented in figure 5.3 for leakage current against the difference in junction depth and EOR damage depth ($x_j - x_{EOR}$). Leakage maximizes when EOR damage is at or near the junction of the material. This comes as little surprise since damage in the space-charge region causes leakage [Ash77, Bul79]. A depletion width of $\sim 730$ Å is calculated assuming SPE regrowth generates a hole concentration of $2 \times 10^{20}$ cm$^{-3}$ in an implant layer whose concentration gradient is assumed to be a step function. Since the furthest the damage lies away from the junction is $\sim 600$ Å for these samples, the damage always lies in the depletion region. The scatter in the data for a fixed $x_j - x_{EOR}$ suggests there are other factors involved in leakage currents than simply $x_j$ relative to $x_{EOR}$.

Several other variables affecting leakage are explored to determine if correlations exist. Simply comparing defect data directly to leakage current does not produce obvious trends but fixing $x_j - x_{EOR}$ allows relationships to be extracted. When values of $x_j - x_{EOR}$ are within $\sim 30$ Å of each other they are considered about equal and an average $x_j - x_{EOR}$ is calculated. Figures 5.4 and 5.5 plot interstitial density, average maximum defect length, and average minimum defect length versus the leakage current. Because defects may not be perfect circles they will have a maximum and minimum “diameter”, which is averaged over all defects counted. From these plots it cannot be said for certain that interstitial density and defect “diameter” have a direct effect on the leakage current of the material. If the defect density and line length are now related to the leakage current the results are shown in figures 5.6 and 5.7. These figures offer evidence of increasing defect density and line length raising the leakage current. When $x_j - x_{EOR}$ is close to 0, the sensitivity to
changes in defect density or line length generates smaller changes in leakage. It makes more sense that defect density and line length would better correlate to leakage if it is indeed the strained bonds at the edges of the defect initiating leakage.

**Discussion**

Only considering an amorphous layer deep enough to activate the entire dopant profile leads to serious consequences in the quality of the electrical junction. Figure 5.3 supplies evidence that damage at the junction maximizes leakage, producing material unsuitable for devices, but these results need to be carefully approached. The depth in the SIMS profile where boron concentrations reach $1 \times 10^{18}$ cm$^{-3}$ defines a junction for this work but is not absolutely the electrical junction. In fact, studies exploiting stripping Hall and spreading resistance profiling (SRP) reveal the tail end of a boron profile remains inactive after the SPE anneal if the implant reaches beyond the $\alpha/c$ interface [Lin02, Tsa79a]. Facts also point to diffusion of boron to the EOR damage but it remains inactive [Lin02]. It is interesting to note how electrical junction depth is controlled by the depth of the amorphous layer in SPE regrown materials, making energy contamination during the dopant implant step less significant for USJ materials. Pushing the tail of the boron profile deeper into the crystalline substrate reduces junction leakage and should retain a shallower junction at the same time. This idea is vital to contemplate when designing transistors from a SPE regrowth process. While this understanding of leakage due to damage proximity to junctions is certainly helpful, the acceptable amount needs to be taken into account also.

To gain further insight into observed values of leakage requires knowledge of device limitations. The ITRS reports off-state leakage levels permissible for future MOSFETs for high performance, low operating power (LOP), and low standby power (LSTP)
devices [ITR01]. The off-state leakage values in the roadmap are given in current/distance so comparison to data from the above experiment calls for ITRS values to be divided by the printed gate length for estimation of leakage current per area. The ITRS values are for total off-state leakage from all contributing sources of leakage (e.g. junction, gate, etc.) so the experimental data should only be a small fraction of the ITRS values. The experimental data are plotted with the ITRS numbers (using the larger junction depths for each technology node) in figure 5.8. The SPE regrown material has suitable characteristics for high performance devices and, if tailored correctly, may meet the needs of LOP devices, but LSTP devices appear out of reach currently. The high leakage from SPE grown devices would be draining to battery operated devices without further optimizing material performance. Since increased defect density raises leakage currents, it seems highly likely that the use of larger preamorphizing species, such as germanium, would benefit leakage characteristics by reducing defect densities [Cla01].

**Summary**

A study of leakage characteristics is SPE regrown materials is presented here since it is crucial to contemplate when designing materials for future devices. Leaving the tail end of a boron implant at or near the original α/c interface leads to high leakage currents, thus needs to be avoided. It is not obvious if interstitial density and defect “diameter” change leakage levels but increasing defect density and line length augments leakage currents for these SPE regrown materials. Material design should be examined carefully for its application. Keeping the junction away from EOR damage improves leakage and pushing the boron tail beyond the α/c interface deserves thought. Further investigation into true electrical junctions is essential for a full understanding. Finally, SPE regrown
materials meet some ITRS values but more optimization work for the strict LSTP standards is desirable.
Table 5.1. Experimental implant conditions for junction leakage current study.

<table>
<thead>
<tr>
<th>Preamorphization Implant (keV)</th>
<th>PLAD Energy (kV)</th>
<th>PLAD Dose (cm$^2$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>5</td>
<td>0.6</td>
<td>5x10$^{15}$</td>
</tr>
<tr>
<td>5</td>
<td>2</td>
<td>5x10$^{15}$</td>
</tr>
<tr>
<td>10</td>
<td>2</td>
<td>5x10$^{15}$</td>
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<td>10</td>
<td>5</td>
<td>5x10$^{15}$</td>
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<tr>
<td>10</td>
<td>5</td>
<td>2x10$^{16}$</td>
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<tr>
<td>20</td>
<td>0.6</td>
<td>5x10$^{15}$</td>
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<td>5x10$^{15}$</td>
</tr>
<tr>
<td>30</td>
<td>5</td>
<td>2x10$^{16}$</td>
</tr>
</tbody>
</table>
Figure 5.1. Schematic of experimental set-up showing a) junction shallower than EOR damage, b) junction at EOR damage, and c) junction deeper than EOR damage.
Figure 5.2. Depth of damage at EOR versus the amorphous layer thickness showing damage layer depth moving farther from amorphous/crystalline as implant energy is increased.
Figure 5.3. Leakage current as a function of $x_j - x_{EOR}$. Leakage is maximized when the tail of the boron implant is at or near the damage layer.
Figure 5.4. Effect of defect diameter on the leakage current for a) the average minimum dimension of defects and b) the average maximum dimension of defects.
Figure 5.5. Impact of interstitial density on leakage current. It is not clear if increased interstitial density raises the leakage current.
Figure 5.6. Increasing defect density amplifies the junction leakage current when fixing $x_j-x_{EOR}$. 
Figure 5.7. Line length increases the leakage current for junctions at fixed $x_j - x_{EOR}$.
Figure 5.8. ITRS estimated leakage requirements plotted along with the data from this experiment. High performance and possibly LOP standards are met but LSTP requirements are tougher to reach.
CHAPTER 6
DETERMINING ACTIVE BORON CONCENTRATIONS FROM SOLID PHASE EPITAXIALLY REGROWN SILICON

Introduction

Solid phase epitaxial (SPE) growth processes are receiving renewed interest due to demanding requirements for future metal oxide semiconductor field effect transistor (MOSFET) devices. Using SPE allows active dopant concentrations to reach higher levels while exhibiting little diffusion at lower temperatures. The International Technology Roadmap for Semiconductors (ITRS) [ITR01] constraints displayed in chapter 1, figure 1.2 convey the need for high dopant activation since junction depths are considerably shallow. The estimates of active boron concentrations that reach sheet resistances in figure 1.2 are shown in figure 1.3 and indicate necessary boron active levels in the $10^{20}$ cm$^{-3}$ range. Integrating an SPE process is not too difficult considering existing equipment suffices and complimentary MOSFET (CMOS) devices using SPE demonstrate sufficient performance [Hir89, Kum01, Kum02, Liu01, Liu02, Mit96, Miy00, Tsu99]. Current SPE research [Fel02, Har98, Lin00, Lin01, Lin02, Osb96] into generating ultra-shallow junction (USJ) materials promises to reach sheet resistance and junction depth specified by the ITRS but there is not much information on attainable active levels for boron in silicon.

To determine active boron levels requires process optimization and characterization using an electrical technique such as Hall effect or spreading resistance profiling (SRP). The SPE follows an Arrhenius relationship so it is beneficial to examine ranges of anneal
temperatures and times as well as initial boron levels in silicon. The impact of amorphizing species also plays a role in optimizing electrical properties. Choosing appropriate boron implant energy becomes an important concern in the electrical characterization. The work in this chapter examines SPE regrown material using Hall effect making USJ material characterization somewhat difficult at times. Deeper implants are chosen to reduce risks of contacts punching through the implant layer and allow accurate secondary ion mass spectrometry (SIMS) measurements. The deeper junctions also reduce the chance of probes penetrating beyond the implant layer when taking sheet resistance measurements with four-point probe. These investigational considerations make it easier to determine active boron concentrations but do not produce junctions shallow enough for modern technology requirements.

Experimental

Experimental implant conditions are divided into two sets based on the preamorphization conditions. Implants into 200 mm, n-type <100> silicon wafers of 9-10 Ω·cm resistivity with either 30 keV, 1x10^{15} cm^{-2} Si^+ or 60 keV, 1x10^{16} cm^{-2} Ge^+ spawns amorphous layers ~680 Å or ~1180 Å respectively. Each preamorphized set is then implanted with 5 keV B^+ in the dose range of 5x10^{14}–8x10^{15} cm^{-2}. A wafer at each preamorphization condition is set aside without boron implantation. For boron dose of 5x10^{14} cm^{-2} an implant current of 7 mA is used while implant doses of 1x10^{15} cm^{-2} and higher use 14 mA. Boron junction depths reach ~700-1000 Å for depths defined at boron concentration of 1x10^{18} cm^{-3}. Samples are cut from the wafers by dicing the wafers into 1.5 cm x 1.5 cm pieces.

Preparation of samples for characterization includes annealing at low temperatures (500-650 °C) for various times. If the anneal is less than 10 minutes it is performed in an
AG Associates Heat Pulse 4100 rapid thermal anneal (RTA) system while anneals greater than 10 minutes are done in a quartz tube furnace. The RTA anneal ramp rate is set at 50 °C/sec and temperature profiles show overshoot of ≤10 °C. All anneals are under ultra-high purity nitrogen (N₂) purge. The regrowth process follows an Arrhenius trend so anneal times are adjusted according to amorphous layer depth and anneal temperature.

Organics and other debris are removed with acetone before samples are processed and characterized. Amorphous layer thickness is measured using a J A Woollam, M-88 variable angle spectroscopic ellipsometer (VASE) at angles of 65, 70, and 75°. SIMS measures boron concentration profiles utilizing an O₂ source beam. Transmission electron microscopy (TEM) on a JEOL 200CX examines several samples in plan-view. Imaging is taken using g₁₂₀ weak beam dark field (WBDF) with g₃g diffraction conditions. The quantitative TEM (QTEM) methods used to analyze the images are described elsewhere [Bha97]. Electrical characterization is comprised of both four-point probe and Hall effect. The four-point system is by Jandel Engineering Ltd. with tungsten-carbide probes that are in-line and spread 1.016 mm apart. The force applied to samples is 60+ g per probe for probes of 100 µm radius. Current is fixed at either 438.02 µA or 453.24 µA but several samples are measured at various currents to confirm ohmic contacting. The Hall system is by MMR Technologies and includes the M-50 bench top electromagnet, MPS-50 programmable magnet power supply, K-20 programmable temperature controller, and H-50 Hall, van der Pauw controller. The Hall and van der Pauw software (Hall and van der Pauw Measurement System Software Package v. 2.0) used by the Hall system chooses a current based on 85% of the minimum current between any pair of contacts when the potential is maintained at 1.5 V. Indium metal pressed onto
the corners of samples form ohmic contacts. Field is generally fixed at 3000 G and measurements are taken at room temperature and 200 K. For select samples the current, field, and temperature are varied to further explore anomalous results. The Hall scattering factor is assumed to be $r=0.7$ again due to the high active hole concentrations.

**Results**

**Silicon Preamorphized**

Finding information on regrowth behavior of undoped silicon allows comparison to previous findings discussed in chapter 2. The evaluated results help determine if similar behavior exists, and makes it easier to predict times to complete regrowth. Samples implanted with only 30 keV, $1 \times 10^{15}$ cm$^{-2}$ Si$^+$ (no boron) are annealed at 500-650 °C for various times and amorphous layers are measured using VASE. Figures 6.1 a) and b) plot the remaining amorphous layer thickness versus anneal time for these samples. Data fitting indicates highly linear trends so regrowth velocity is constant. The initial amorphous thickness is omitted due to inaccuracies in VASE measurements from rough interfaces because of incomplete amorphization at the amorphous/crystalline ($\alpha$/c) interface. This rough interface may also exhibit an initial regrowth rate different from the steady-state regrowth rate. Using the regrowth velocities allows an Arrhenius plot to be constructed as in figure 6.2. The slope indicates an activation energy of 2.51 eV and pre-exponential factor of $1.67 \times 10^{17}$ Å/min. These results verify the behavior of SPE regrown silicon discussed in chapter 2. Now examining regrowth with the presence of boron would predict an increased regrowth rate and is displayed in figure 6.3. The initial amorphous layer thickness is determined after annealing samples at 400 °C for 60 minutes to smoothen the $\alpha$/c interface. Figure 6.3 plots the amorphous layer thickness versus anneal time at 500 °C for all the boron implant doses demonstrating the increased
regrowth rate with increasing boron dose which is maximized at $\sim 2-4 \times 10^{15}$ cm$^{-2}$ boron, followed by a slight slow down at higher doses. Highest regrowth enhancements are observed when boron concentrations are $\sim 4 \times 10^{20}$ cm$^{-3}$ [Ker84], which is the peak concentration for the $1 \times 10^{15}$ cm$^{-2}$ boron implant. It makes sense then that the medium doses lead to the fastest regrowth because deeper layers with boron concentrations in the $10^{20}$ cm$^{-3}$ range are formed. Regrowth slows down significantly and stops around $\sim 4 \times 10^{21}$ cm$^{-3}$ boron concentrations [Ker84], which is approximately the peak concentration of the $8 \times 10^{15}$ cm$^{-2}$ boron implant. Keeping the regrowth rate adjustments in mind allows estimates of time to complete regrowth at other temperatures by making use of the ratio of regrowth velocities. The Arrhenius relation of regrowth velocity permits regrowth times to be calculated when the activation energy is known. The estimated times ensure complete regrowth of the amorphous layers, which would be required in an integrated circuit (IC) process.

Further study into the electrical characteristics reveals some interesting but sometimes confusing results. Sheet resistance measurements plotted in figures 6.4 a)-d) for varying dose and at various times yield an optimized boron dose of $\sim 2-4 \times 10^{15}$ cm$^{-2}$, which is where the fastest regrowth occurs. For the highest boron dose, regrowth may not be complete, but the higher temperatures appear to help reduce the sheet resistance of these samples more so than the lower temperatures. The sheet resistance versus anneal temperature for times of 45 minutes, 3 minutes, 45 seconds, and 4 seconds at temperatures of 500, 550, 600, and 650 °C respectively is plotted in figure 6.5 to show the slight advantage of moving to higher anneal temperatures. Whether the advantage comes from dopant activation or lattice repair is examined later. Samples annealed at 600 °C for
longer times are shown in figure 6.6 with the 600 °C/45 second data. Clearly the much longer times have little effect on sheet resistance suggesting little diffusion and minimal impact on the electrical properties of the material. Deactivation studies would then need to be conducted at higher temperatures, however, this study focuses more on the activation characteristics.

While sheet resistance measurements from the four-point probe are easily obtained and interpretable, the Hall measurements presented some problems. Typical room temperature measurements produced electrons as primary carriers so various parameters are changed to establish why. Figures 6.7 and 6.8 vary the current and magnetic field respectively for measurements of active dose for a sample with 1x10^{15} cm^{-2} boron annealed at 650 °C for 4 seconds. Figure 6.8 also fixes several sample temperatures while varying magnetic field. For the most part, the sample active dose is not affected by either current or magnetic field, but temperature seems to cause changes in whether the sample measures holes or electrons. Figures 6.9 a) and b) illustrate this better by plotting the active dose and Hall mobility versus sample temperature during Hall measurement. This kind of behavior is seen in HgCdTe films [Gol86, Lou84, Zem87], but these films are also affected by magnetic field. It is also known that defects in the space-charge region cause leakage current to rise [Ash77, Bul79], leading to the possibility of substrate and implant layer measurement coincidently. A more thorough investigation into “leaky” junction, silicon-based materials by Larrabee and Thurber [Lar80] applies contacting to both the implant layer and substrate since a resistance still exists between the layers. This technique is not employed here due to complexities of the set-up. More recent work [Jai03] of electron measurements for boron implants into silicon confirms the potential
for trouble in “leaky” measurements. For the sample in figure 6.9 the transition from electron to hole occurs at ~250 K but this transition may occur at slightly higher or lower temperatures depending on defect proximity in the space-charge region.

The event of electron measurements puts serious doubt into reliability and accuracy of Hall measurements as well as four-point probe values. Therefore a sample with boron implant of $2 \times 10^{15}$ cm$^{-2}$ is annealed at 500 °C for 45 minutes and measured by four-point probe (at room temperature), room temperature Hall, and 200 K Hall measurement. The sample is then annealed at 900 °C for 30 minutes and measured again by four-point probe, Hall, and low temperature Hall. The results are given in table 6.1 and show the four-point probe measurements match low temperature Hall after the SPE anneal. After the 900 °C/30 minute anneal the leakage no longer becomes an issue, so room and low temperature Hall match four-point probe data. Thus the four-point probe measurements are reasonable while only low temperature Hall data are acceptable after low temperature SPE annealing. The temperature of 200 K for Hall measurements is chosen to freeze out contribution to leakage current from end-of-range (EOR) damage but is not cold enough to freeze out extrinsic hole carriers. Then plotting the active dose versus implant dose in figure 6.10 describes how a maximum active dose is achieved from SPE and levels off for higher implant dose. The hole mobility plot in figure 6.11 verifies the sheet resistance results by conveying decreased mobility at higher doses. The percent activation is determined using the boron dose from SIMS data and results are shown in figure 6.12. As expected, the percentage of active boron goes down as implant dose is increased. The error in the Hall measurements is determined by calculating the standard deviation in the measurement of 10 samples implanted with 30 keV, $1 \times 10^{15}$ cm$^{-2}$ Si$^+$ and 5 keV, $4 \times 10^{15}$
cm$^2$ B$^+$ then annealing at 500 °C for 45 minutes. Further manipulation of the activation data is discussed later.

**Germanium Preamorphized**

Similar steps are taken to study samples preamorphized with germanium rather than the silicon preamorphization just discussed. To begin, regrowth studies verified how germanium has little effect on SPE regrowth. Figures 6.13 a) and b) illustrate the amorphous layer versus anneal time while figure 6.14 is the Arrhenius plot with slope of 2.46 eV and pre-exponential factor of 5.92x10$^{16}$ Å/min. Boron implanted sample regrowth characteristics at 550 °C in figure 6.15 display similar trends of maximum regrowth velocity at ~2-4x10$^{15}$ cm$^{-2}$ boron dose. Thus these samples behave similar to the silicon preamorphized samples but have a deeper initial amorphous layer due to the 1x10$^{16}$ cm$^2$ Ge$^+$ dose used. This means the electrical characteristics behave differently due to the EOR damage being pushed deeper.

Electrical measurements consist of four-point probe and low temperature Hall once again. Various samples measure an accurate active dose of holes using room temperature Hall but since others do not and all the silicon preamorphized samples measure electrons, 200 K is chosen for sample measurement again. Sheet resistance from four-point probe measures accurately at room temperature again and results versus implant dose for 500-650 °C anneals are shown in figures 6.16 a)-d). It is interesting to note the optimum boron dose has now shifted to ~4-6x10$^{15}$ cm$^{-2}$. The sheet resistance as a function of anneal temperature displays the similar behavior to that of silicon preamorphized samples in figure 6.17 (increasing anneal temperature decreases sheet resistance). The samples of interest are annealed at 500, 550, 600, and 650 °C for 2.5 hours, 16 minutes, 2.5 minutes, and 30 seconds respectively. These samples are used for Hall measurements also. Active
dose maximizes at an implant dose of $\sim 4 \times 10^{15} \text{ cm}^{-2}$ boron and remains constant at higher implant dose as seen in figure 6.18. Hole mobility plotted in figure 6.19 shows the same general decreasing mobility with increased boron implant dose, but active doses are slightly lower for germanium preamorphized than silicon preamorphized samples. Finally the percent activation plotted with implant dose in figure 6.20 deduces the expected trend of decreasing percentage of active boron with increasing boron implant dose, confirming that much of the boron remains off lattice sites at higher doses. Error in germanium preamorphized samples is estimated by the standard deviation in 10 measured samples implanted with 5 keV, $6 \times 10^{15} \text{ cm}^{-2} \text{ B}^+$ and annealed at 500 °C for 2.5 hours.

**Discussion**

Results from electrical measurements seem ambiguous at times, however interpreting data correctly allows determination of accurate material properties. Sheet resistance values are resolved from both Hall and four-point probe measurements with matching results for Hall samples tested at 200 K and room temperature four-point probe samples. Therefore, sheet resistance values are largely obtained from the four-point probe system since it is much quicker and easier. The room temperature Hall measurements may yield unreasonably low sheet resistance values and electrons as the primary carrier due to large junction leakage currents. These leakage currents presumably allow the measurement of the n-type substrate in parallel with the active implant layer. Why the four-point probe measures accurate results at room temperature while Hall data varies is not fully understood but may be a result of the longer path the forced current travels for Hall measurement as compared to four-point measurements ($\sim 20 \text{ mm}$ for Hall versus $\sim 3 \text{ mm}$ for four-point probe).
Sheet resistance data compared to Hall data gives way to remarkable behavior in the SPE regrown boron in silicon. Tradeoffs between reaching minimum sheet resistance and fully activating the boron dose exist owing to the Gaussian nature of the dopant implant profile. The lowest doses are fully or nearly fully active while the $2-4 \times 10^{15}$ cm$^{-2}$ boron doses and $4-6 \times 10^{15}$ cm$^{-2}$ boron doses minimize the sheet resistance for silicon and germanium preamorphized samples respectively. This optimum occurs because a thicker layer with maximum active boron concentration is achieved as dose is increased rather than a boron concentration profile that has just reached the maximum hole concentration. As boron concentration is increased though, the excess boron remains inactive and contributes scattering sites thus increasing sheet resistance.

The differences seen in sheet resistance of silicon and germanium preamorphized samples come from the combined effects of boron activation and hole mobility. Comparing figures 6.10 and 6.18 illustrates the benefit of silicon preamorphization to obtain slightly higher active doses of boron. In similar respects, figures 6.11 and 6.19 show the slight advantage of using silicon preamorphization in mobility. It is suspected that germanium would act as an additional scattering site due to its larger size within the silicon matrix. The presence of germanium also lessens the amount of active boron despite hope that strain compensation between boron and germanium would elevate boron activation. This suggests germanium-boron complexes may be forming and lowering the amount of born available to sit on lattice sites. Another worthy note is the reduced mobility of germanium preamorphized samples at lower doses of boron compared to silicon preamorphized counterparts while higher boron doses show more similar mobilities for both preamorphization conditions. Since the mobility of
germanium preamorphized samples does not drop as much as silicon preamorphized samples when boron dose is high it could explain the shift to higher boron doses for optimum sheet resistance. Hence there appears to be a slight gain in mobility for samples that are better strain compensated. The germanium could be reducing distortions in the lattice caused by boron.

More information about realized active boron concentrations is now needed to verify silicon preamorphization generates higher hole concentrations and that these concentrations are sufficient to meet needs of future devices. The approach consists of adjusting the as-implanted boron SIMS concentration profiles (boron is assumed to move little at these low anneal temperatures) by finding a maximum boron concentration where the dose integrated from the adjusted profile matches the active dose from the Hall system. This is demonstrated in figure 6.21 for a sample preamorphized with silicon and implanted with $2 \times 10^{15}$ cm$^{-2}$ boron, then annealed at 650 °C for 4 seconds. The computer program given in appendix C calculates a boron concentration where the dose under the curve equals that of the dose from 200 K measurements on the Hall system (the rest of the boron above the concentration is assumed inactive). This adjusted concentration profile is then thought of as a hole concentration profile for the active implant layer and used to calculate a theoretical sheet Hall coefficient, $R_{Hs}$, (the Hall coefficient, $R_H$, is derived in appendix B but $R_{Hs}$ is simply $R_H/t$ where $t$ is the thickness of the measured layer) given by:

$$R_{Hs} = \frac{\int_0^t q n_p(z) \mu_p^2(z) dz}{\left( q \int_0^t n_p(z) \mu_p(z) dz \right)^2}$$  \hspace{1cm} (6.1)
where $n_p(z)$ is the hole concentration at depth of $z$, $\mu_p(z)$ is the hole mobility at concentration $n_p(z)$, and $q = 1.6 \times 10^{-19}$ C. This requires a model for the mobility as a function of hole concentration. Several models exist [Aro82, Kla92, Li78, Mas83], but the model by Masetti et al. provides sufficient theoretical mobility values for the concentration ranges used in these experiments. Applying the computer program to the Hall data allows the maximum achievable active boron concentrations to be determined and are shown versus boron implant dose in figures 6.22 and 6.23 for silicon and germanium preamorphized samples respectively. These are the same Hall samples used in figures 6.10 and 6.18. These figures demonstrate boron activation clearly reaches $\sim 2-3 \times 10^{20}$ cm$^{-3}$ in silicon preamorphized samples and $\sim 1-2 \times 10^{20}$ cm$^{-3}$ in germanium preamorphized samples. These emerge as the highest levels of active boron that can be acquired in these SPE experiments and are sufficiently high to reach several generations of ITRS requirements given in figure 1.3. The SPE process may have a slight boron activation effect from the anneal temperature range, which is better illustrated in figures 6.24 (silicon preamorphized) and 6.25 (germanium preamorphized) for the same data in figures 6.22 and 6.23. The advantage to sheet resistance at higher anneal temperatures must come from a combination of better lattice repair (increased mobility) and slightly higher active boron concentrations. For the silicon preamorphized samples in figure 6.24, there appears to be some increase in active boron levels, but this is not the case for germanium preamorphized samples shown in figure 6.25. As mentioned, the hole concentration profile is used to determine a theoretical sheet Hall coefficient, which is then compared to experimental values. The difference between theoretical and experimental data is $\leq 10\%$, which is in excellent agreement and suggests the data in
figures 6.22-6.25 is an accurate account of active boron levels attainable by SPE regrowth.

An attempt to model the Hall results measured at room temperature is met with much less success. Since the junctions are very leaky in these samples, a two-layer approach to modeling the sheet Hall coefficient is also derived in appendix B. It is given by:

\[
R_{hs} = \frac{\int_0^{t_1} q n_p(z) \mu_n^2 (z) dz - q n_n \mu_n^2 t_2}{q^2 \left( \int_0^{t_1} n_p(z) \mu_p(z) dz + n_n \mu_n t_2 \right)^2}
\]

where the new variables added are \(n_n\), the doped substrate concentration, \(\mu_n\), the mobility of the substrate, and \(t_1\) and \(t_2\) are the thickness of the implant layer and substrate respectively. Substrate thickness is calculated to compare the known thickness by using the hole concentration profile determined by the computer program along with known substrate mobility and concentration. Some sample’s data converges onto an imaginary substrate thickness so the room temperature Hall data cannot be used to calculate hole concentration profiles using a fixed substrate thickness. The biggest contribution to error in room temperature measured Hall coefficients probably comes from the unaccounted resistance between the implant layer and substrate, which is addressed by Larrabee and Thurber [Lar80]. If the layers are simply modeled as parallel resistors and sheet resistances are used to calculate the thickness of an n-type layer needed to match the value measured by room temperature Hall it is found the substrate would need to be \(\sim 350-550 \mu m\) thick. Since the substrates are sufficiently thick (\(\sim 725 \mu m\)) this provides further evidence the samples are experiencing junction leakage issues when measured with room temperature Hall.
A curious observation in the room temperature Hall measurements of the germanium preamorphized samples led to further study of SPE regrown samples. For many germanium preamorphized samples, holes are measured except for samples regrown at 600 ºC. This set of samples displayed characteristics of electron carriers or impossible hole concentrations. Clearly these samples are subject to greater junction leakage currents and they have similar junction proximity to EOR damage as other regrown, germanium preamorphized samples so EOR defect population is apparently varied significantly even at low anneal temperatures. Samples silicon preamorphized with a boron dose of $2 \times 10^{15}$ cm$^{-2}$ and germanium preamorphized with a boron dose of $6 \times 10^{15}$ cm$^{-2}$ are annealed at 500-650 ºC for the same times as the Hall samples. A set of samples using the material from the ultra low energy (ULE) experiment in chapter 4 are also prepared for a silicon preamorphization of 2.5 keV, $1 \times 10^{15}$ cm$^{-2}$ and boron implant of 0.5 keV, $1 \times 10^{15}$ cm$^{-2}$. These samples are annealed in the quartz tube furnace at 500-650 ºC for 4 min (all temperatures). The QTEM from all samples is shown in figures 6.26-6.28 for silicon, germanium, and ULE experiment samples respectively. The defect density and line length versus anneal temperature show maximums in the 550-600 ºC range as expected since these samples have the greatest leakage problems. From chapter 5 it is concluded that higher defect densities and line lengths increase leakage currents, corroborating this finding. These results are also consistent with other findings where 500-600 ºC pre-anneals followed by a 950 ºC spike anneal maximizes the defect density [Cam02]. Certainly this creates additional consideration when developing the SPE technique if it is to be used in a device manufacturing process.
Summary

Introducing an SPE process into IC manufacturing is possible, but many factors need to be addressed. The tradeoffs between boron implant dose and percent activation show the highest percentage of activation will not yield the lowest sheet resistance. Furthermore, high doses of germanium as a preamorphization agent provides little benefit to electrical properties. Activation measurements using Hall systems are sometimes difficult because of junction leakage inherent to SPE regrown samples. The problems are overcome by cooling samples down, which eliminates any leakage paths at the junction. Despite much of the boron remaining inactive at higher doses, SPE achieves hole concentrations of $\sim 2-3\times10^{20}$ cm$^{-3}$ for silicon preamorphization and $\sim 1-2\times10^{20}$ cm$^{-3}$ for germanium preamorphization. These concentrations are sufficient to meet ITRS requirements for the next few technology nodes. Finally, defect analysis shows defect population is maximized in the 550-600 °C annealing range. These important findings allow a better understanding to the SPE process and give way to ideas for future experimental design.
Figure 6.1. Regrowth characteristics of undoped silicon preamorphized with 30 keV, 1x10^{15} \text{cm}^{-2} \text{ silicon and annealed in the temperature range of 500-650 °C. Data for anneals are split into a) 500 & 550 °C and b) 600 & 650 °C.
Figure 6.2. Arrhenius plot of silicon regrowth preamorphized with 30 keV, $1\times10^{15}$ cm$^{-2}$ silicon.
Figure 6.3. Regrowth of 30 keV, 1x10^{15} cm^{-2} silicon preamorphized samples implanted with various 5 keV boron doses at 500 °C showing maximum regrowth velocity for middle boron doses.
Boron Implant Dose ($\text{cm}^{-2}$)

Sheet Resistance ($\Omega/$sq)

500 °C

- 30 min
- 35 min
- 40 min
- 45 min

SPE regrowth incomplete

regrowth still occurring

550 °C

- 3 min
- 3.5 min
- 4 min
- 4.5 min
- 5 min
- 5.5 min
Figure 6.4. Sheet resistance versus implant dose for 30 keV, $1 \times 10^{15}$ cm$^{-2}$ silicon preamorphized samples indicating optimized boron dose of $\sim 2-4 \times 10^{15}$ cm$^{-2}$ for anneal temperatures of a) 500 °C, b) 550 °C, c) 600 °C, and d) 650 °C. Samples are implanted with 5 keV B\textsuperscript{+} and various anneal times are shown.
Figure 6.5. Sheet resistance as a function of anneal temperature for 30 keV, $1 \times 10^{15}$ cm$^{-2}$ silicon preamorphized samples at anneal times of 45 minutes, 3 minutes, 45 seconds, and 4 seconds for 500, 550, 600, and 650 °C respectively.
Figure 6.6. Extended anneal times at 600 °C show little effect on the sheet resistance values suggesting deactivation at low temperatures is not a problem. Preamorphized with 30 keV, $1 \times 10^{15}$ cm$^{-2}$ Si$^+$ followed by 5 keV B$^+$ implants.
Figure 6.7. Active dose from room temperature Hall measurement on a sample preamorphized with 30 keV, 1x10^{15} \text{ cm}^{-2} \text{ Si}^{+} and implanted with 5keV, 1x10^{15} \text{ cm}^{-2} boron annealed at 650 \degree C for 4 seconds. Hall measures electron carriers with little effect from current except at lower values where noise may become an issue.
Figure 6.8. Magnetic field strength has little effect on active dose measurements for
sample with 30 keV, $1 \times 10^{15}$ cm$^{-2}$ Si$^+$ and 5 keV, $1 \times 10^{15}$ cm$^{-2}$ boron annealed at 650 °C
for 4 seconds. Lower temperatures measure holes while higher temperatures measure
electrons.
Figure 6.9. Sample temperature during Hall measurement plays a key role in determining whether holes or electrons are measured. For the sample with 30 keV, $1 \times 10^{15}$ cm$^{-2}$ Si$^+$ and 5 keV, $1 \times 10^{15}$ cm$^{-2}$ boron annealed at 650°C for 4 seconds the transition from between electron and hole measurement occurs between 250 and 260 K. Active dose is shown in a) while Hall mobility is shown in b) as a function of sample temperature.
Table 6.1. Four-point probe and Hall results compared to determine accuracy of measurements.

<table>
<thead>
<tr>
<th>Sample Temperature</th>
<th>Measurement System</th>
<th>900 °C/30 min anneal?</th>
<th>Sheet Resistance (Ω/sq)</th>
<th>Carrier type</th>
</tr>
</thead>
<tbody>
<tr>
<td>room temp.</td>
<td>four-point</td>
<td>no</td>
<td>141</td>
<td>electron</td>
</tr>
<tr>
<td>room temp.</td>
<td>Hall</td>
<td>no</td>
<td>80</td>
<td>electron</td>
</tr>
<tr>
<td>200 K</td>
<td>Hall</td>
<td>no</td>
<td>141</td>
<td>hole</td>
</tr>
<tr>
<td>room temp.</td>
<td>four-point</td>
<td>yes</td>
<td>135</td>
<td>hole</td>
</tr>
<tr>
<td>room temp.</td>
<td>Hall</td>
<td>yes</td>
<td>125</td>
<td>hole</td>
</tr>
<tr>
<td>200 K</td>
<td>Hall</td>
<td>yes</td>
<td>128</td>
<td>hole</td>
</tr>
</tbody>
</table>
Figure 6.10. Active boron dose plotted against implant boron dose for 30 keV, $1 \times 10^{15}$ cm$^{-2}$ Si$^+$ preamorphized samples showing maximum active boron dose of $\sim 1 \times 10^{15}$ cm$^{-2}$ at implant dose of $2 \times 10^{15}$ cm$^{-2}$ and remains about constant for higher implant dose.
Figure 6.11. Hole mobility for 30 keV, $1 \times 10^{15}$ cm$^{-2}$ Si$^+$ preamorphized samples showing general decreasing trend as 5 keV boron implant dose is increased.
Figure 6.12. Percent active boron decreases as implant dose is increased. Samples are preamorphized with 30 keV, 1x10^{15} \text{ cm}^{-2} \text{ Si}^+ and implanted with 5 keV boron.
Figure 6.13. Regrowth characteristics of undoped silicon preamorphized with 60 keV, 1x10^{16} cm^{-2} germanium and annealed in the temperature range of 500-650 °C. Data for anneals are split into a) 500 & 550 °C and b) 600 & 650 °C.
Figure 6.14. Arrhenius plot of silicon regrowth preamorphized with 60 keV, $1 \times 10^{16}$ cm$^{-2}$ germanium.
Figure 6.15. Regrowth of 60 keV, $1 \times 10^{16}$ cm$^{-2}$ germanium preamorphized samples implanted with various 5 keV boron doses at 550 $^\circ$C showing maximum regrowth velocity for middle boron doses.
Figure 6.16. Sheet resistance versus boron implant dose for 60 keV, 1x10^{16} \text{cm}^{-2} germanium preamorphized samples indicating optimized boron dose of \(\sim 4-6 \times 10^{15} \text{cm}^{-2}\) for anneal temperatures of a) 500 °C, b) 550 °C, c) 600 °C, and d) 650 °C. Samples are implanted with 5 keV B\(^{+}\) and various anneal times are shown.
Figure 6.17. Sheet resistance as a function of anneal temperature for 60 keV, $1 \times 10^{16}$ cm$^{-2}$ germanium preamorphized samples with 5 keV boron at anneal times of 2.5 hours, 16 minutes, 2.5 minutes, and 30 seconds for 500, 550, 600, and 650 $^\circ$C respectively.
Figure 6.18. Active boron dose plotted against boron implant dose showing maximum active boron dose of \(~1 \times 10^{15}\) cm\(^{-2}\) at implant dose of \(4 \times 10^{15}\) cm\(^{-2}\) and remains constant for higher implant dose. Samples are preamorphized with 60 keV, \(1 \times 10^{16}\) cm\(^{-2}\) Ge\(^+\) and implanted with 5 keV B\(^+\).
Figure 6.19. Hole mobility showing general decreasing trend as 5 keV boron implant dose is increased. Samples are preamorphized with 60 keV, 1x10^{16} cm^{-2} Ge^{+}.
Figure 6.20. Percent active boron decreases as 5 keV boron implant dose is increased. Samples are preamorphized with 60 keV, $1 \times 10^{16}$ cm$^{-2}$ Ge$^+$. 
Figure 6.21. Sample SIMS profile with adjusted hole concentration profile used for modeling of sheet Hall coefficient. Implant and anneal conditions are noted in the figure.
Figure 6.22. Maximum active boron concentrations versus the boron implant dose for various anneal temperatures and times. Samples are preamorphized with 30 keV, $1 \times 10^{15}$ cm$^{-2}$ Si$^+$ and implanted with 5 keV B$^+$. 
Figure 6.23. Maximum active boron concentrations versus the boron implant dose for various anneal temperatures and times. Samples are preamorphized with 60 keV, $1 \times 10^{16}$ cm$^{-2}$ Ge$^+$ and implanted with 5 keV B$^+$. 
Figure 6.24. Data from figure 6.22 now plotted versus temperature to show how anneal temperature has little effect.
Figure 6.25. Data from figure 6.23 now plotted versus temperature to show how anneal temperature has little effect.
Figure 6.26. Defect density and line length are maximized in the anneal temperature range of 550-600 °C for samples preamorphized with 30 keV, 1x10^{15} cm^{-2} Si^+ and implanted with 5 keV B^+. Anneal times are 500 °C/45 min, 550 °C/3 min, 600 °C/45 sec, and 650 °C/4 sec.
Figure 6.27. Defect density and line length are maximized in the anneal temperature range of 550-600 °C for samples preamorphized with 60 keV, 1x10^{16} cm^{-2} Ge^{+} and implanted with 5 keV B^{+}. Anneal times are 500 °C/2.5 hr, 550 °C/16 min, 600 °C/2.5 min, and 650 °C/30 sec.
Figure 6.28. Defect density and line length are maximized in the anneal temperature range of 600 °C for samples preamorphized with 2.5 keV, 1x10^{15} cm^{-2} Si^+ and implanted with 0.5 keV B^+. Anneal times are 500-650 °C for 4 minutes.
CHAPTER 7
SUMMARY AND FUTURE WORK

The microelectronics industry continually changes by pushing the limits of material performance and developing techniques for processing. Using approaches such as solid phase epitaxy (SPE) shows promise for extending the life of the metal oxide semiconductor field effect transistor (MOSFET) to future generations of computational devices. The work presented in this dissertation focuses mainly on the electrical characteristics of SPE regrown silicon materials to determine its viability for future technology. As boron doped silicon usually limits MOSFET performance, it is this dopant that is examined. Particularly it is important to study activation traits and relate them to resistivity and mobility. Electrical junction properties play a role in material quality as well, and need to be tailored for optimum performance. A summary of the important findings and suggestions for further study are given here to reiterate the significance of SPE for future material.

Creating high active boron levels in silicon is achievable using current processing techniques at high temperatures, but maintaining shallow junctions is becoming a more difficult task due to transient enhanced diffusion (TED). Low temperature SPE allows ultra-shallow junctions (USJs) to form since diffusion is minimized. Ultra low energy (ULE) experiments (chapter 4) prove boron activates to $>10^{20} \text{ cm}^{-3}$ which is sufficient to reach the next few technology nodes in the International Technology Roadmap for Semiconductors (ITRS). While the sheet resistances of some materials did not meet roadmap requirements, it is possible to further optimize implant conditions to reach ITRS
goals. The sheet resistance versus junction depth requirements from the ITRS are plotted in figure 7.1 for each year up to 2007. Included in this plot are the sheet resistances and junction depths achieved over the years using conventional implants into crystalline or preamorphized silicon and high temperature annealing techniques. As a comparison, results from this SPE work and other SPE work [Fel02] are also plotted for germanium and silicon preamorphized samples. Clearly the SPE regrown samples reach the roadmap, with silicon preamorphized samples performing slightly better than germanium preamorphization. This is why SPE processing cannot be ignored as a possible solution for potential MOSFET production.

Leakage current experiments find junction leakage is maximized when the damage layer is near the tail end of the boron implant. The junction leakage needs to be a small fraction of the overall device leakage so damage should be pushed far beyond the implant tail (deep amorphization) or damage can exist within the implant layer. The latter scheme reduces the total active boron since it will not highly activate at low temperatures in crystalline silicon. Increased defect density and line length raise the junction leakage also. Therefore choosing appropriate preamorphization conditions must be addressed when processing based on an SPE procedure, including both preamorphization depth and species.

Low energy implant experiments provide easier characterization for electrical measurements and greater secondary ion mass spectrometry (SIMS) accuracy but are not directly comparable to today’s requirements for MOSFET devices. However a more fundamental understanding of SPE material property allows insight into the possibility of forming material needed in the future. For ion-implanted dopant, activated using SPE,
there are always tradeoffs for optimizing sheet resistance due to the Gaussian nature of the chemical profile. The SPE regrowth of silicon will activate boron to $\sim 2-3 \times 10^{20} \text{ cm}^{-3}$ so it is necessary to consider the depth of the highly activated layer that creates the desired sheet resistance. The tail, of course, will be active also, and thus determines the junction depth and abruptness, which is of utmost concern for scale-down. Adding too much boron not only increases junction depth, but also leaves inactive impurities that act as scattering sites and lowers the material’s mobility. Picking the appropriate temperature for regrowth is of some concern additionally. Escalated SPE anneal temperatures lower sheet resistance by presumably increasing the mobility and possibly activating higher levels of boron, so apparently low temperature annealing still has significant impact on lattice damage repair. Boron pileup in the end-of-range (EOR) persists (at higher SPE temperatures), pushing boron deeper, but even if boron diffuses in the 600 °C range it does not become active [Lin02], so electrical junction depth is maintained. Utilizing germanium as a preamorphization species has its own tradeoffs in sheet resistance and junction quality too. Germanium increases sheet resistance overall due to additional lattice scattering, but less (EOR) interstitials may form from this heavier element during the preamorphization step. Less EOR damage yields lower junction leakages vital for worthy device properties. The conclusions drawn above clarify some issues preferred for the advancement of MOSFET materials and at the same time perpetuate new questions.

**Future Work**

While this dissertation attempts to explain certain aspects of silicon processing and properties, some new ideas for a more complete understanding come about. Roadblocks along the experimentation process continue to arise, bringing more uncertainties to be
answered. Several suggestions for completeness of this work are offered below as follow-ups for the experiments described above.

- The sheet resistance in the ULE experiment (chapter 4) needs to be optimized by generating implants with lower doses. Several experiments involving USJ materials have already been performed [Lin02, Fel02] and show future technology nodes are met using SPE so it is expected using lower boron doses would fall into future technology requirements as well as providing additional proof of the usefulness of SPE processing.

- Higher anneal temperature experiments will need to be run for deactivation statistics. From the current work, sheet resistance does not change during the anneal periods and is not expected to change with longer anneals $\leq 650^\circ C$ but higher temperatures certainly bring about TED thus changing electrical characteristics and junction depth. This is particularly important to consider for backend processes such as silicidation where temperatures reach $\sim 600-700^\circ C$ [Elk95].

- Further research and development of SIMS and spreading resistance profiling (SRP) techniques leads to better characterization of USJ materials. New equipment is constantly developed for greater accuracy but can be very expensive and thus limits capabilities. In general experiments involving SRP will need to be run varying EOR damage depth relative to the boron tail of the implant. While the junction is typically defined by a boron concentration of $1 \times 10^{18}$ cm$^{-3}$ from the SIMS profile, the actual electrical junction is determined from SRP. The SRP measurements are critical to understanding if the tail-end boron implanted into crystalline silicon has any impact on junction depth after low temperature SPE annealing. This experiment could also involve several implant conditions that fix the difference between the junction depth and EOR damage depth (as shown in chapter 5) so more data points for the effect of defect population etc. on leakage current are gathered. This may involve running implant simulations to determine depths but could also be done by varying the temperature at which the implant is run.

- Further preamorphization studies will include lower doses of germanium and other species such as tin. The lower doses lead to less lattice scattering sites (increasing mobility) and hopefully an improvement in sheet resistance values. The larger preamorphization species also bring about less interstitial damage in EOR benefiting junction properties.

- A final experiment that would be of some interest would be the deposition of amorphous silicon onto crystalline silicon substrates and crystallize the amorphous layers at low temperatures. This experiment has the benefit of having constant concentrations of boron throughout an active layer. The material would also be resistant to diffusion problems until much higher temperatures without the presence of excess interstitials in the EOR.
Figure 7.1. Sheet resistance versus junction depth requirements from the ITRS plotted with results from conventional processing and SPE processing.
APPENDIX A
VARIABLE ANGLE SPECTROSCOPIC ELLIPSMETER THEORY AND APPLICATION

In this section the details of what a spectroscopic ellipsometer measures and its application to theory is presented. A more thorough overview of ellipsometers is left to the reader as there are several references covering this topic [Bar91, Hum93, Woo99]. Some general details of how ellipsometers work will be given for guidance into the discussion of equations relating to the measurements and models. The discussion of equations will be focused into three topics:

1. Equations used to take measurements from samples.
2. Equations used to model reflections from a thick substrate.
3. Equations used to model a thin film layer

The equations shown assume the material is isotropic, which is not always the case, but is a practical assumption.

An outline of the spectroscopic ellipsometer’s functions begin the same as many spectroscopic techniques. A light source generates a beam that is focused onto a sample and reflected into a detector. Of course it is more complex than just shining light onto a sample. A light source that has an acceptable range of wavelengths for the material under study focuses a collimated beam into a polarizer, which in turn generates linearly polarized light. The linearly polarized light is then reflected off the surface of the sample into an analyzer/detector. There are many configurations possible for ellipsometer systems, but the system used in this work is basically set up as mentioned. The linearly
polarized light propagates to the sample and reflects into the analyzer. A plane of incidence is defined by the light leaving the polarizer to the sample and reflecting into the detector (see figure A.1). The electric field vector of the light is inclined some angle, called the azimuth, $\Psi_e$, from the plane of incidence coming from the polarizer and likewise the azimuth for the reflected light, $\Psi_r$, will be some angle from the plane of incidence. This concept is illustrated in figure A.1. Figure A.1 also shows that the electric field can be broken up into two components in the parallel and perpendicular directions denoted by the subscripts “p” and “s” respectively. After the light has reflected off the sample it will become elliptically polarized whereby the wave can now be thought of as having two electric field components perpendicular to one another (and orthogonal to the direction of propagation) and having a phase difference, $\Delta$. It is $\Psi_r$ and $\Delta$ that are characteristic of the material under study and are measured for sample analysis.

**Ellipsometer Measurements**

To begin taking measurements it needs to be understood what is physically happening in the analyzer. The elliptically polarized light from the sample surface will produce a constantly varying intensity of light with time due to the nature of the wave. Therefore a photodetector can simply convert the sinusoidal light signal into a voltage signal that is quantified. The signal will have the form:

$$V(t) = DC + a \cos(2 \omega t) + b \sin(2 \omega t)$$  \hspace{1cm} A.1

It is this fluctuation in intensity that is important to measure and is the key to results.

Much of the theory behind optical phenomena lies in the solutions to Maxwell’s equations for electromagnetic fields. The equations will not be covered in depth here,
instead the important equation derivations will be offered. To begin it is known that the light intensity is proportional to the square of the electric field magnitude, $E_D$:

$$I_D \propto |E_D|^2 \quad \text{A.2}$$

Then using Jones matrix evaluation, an expression for the electric field into the detector is:

$$E_D = \begin{bmatrix} 1 & 0 \\ 0 & 0 \end{bmatrix} \cdot \begin{bmatrix} \cos A & \sin A \\ -\sin A & \cos A \end{bmatrix} \cdot \begin{bmatrix} R_p & 0 \\ 0 & R_s \end{bmatrix} \cdot \begin{bmatrix} \cos \Psi_e & -\sin \Psi_e \\ \sin \Psi_e & \cos \Psi_e \end{bmatrix} \cdot \begin{bmatrix} 1 \\ 0 \end{bmatrix}$$

$$= \begin{bmatrix} R_p \cos \Psi_e \cos A + R_s \sin \Psi_e \sin A \\ 0 \end{bmatrix}$$

where $R_p$ and $R_s$ are the complex Fresnel reflection coefficients from the p and s directions, respectively, and $A$ is the angle between the analyzer exit and plane of incidence. It should also be noted that $\Psi_e$ will be known. Then inserting into A.2 using the complex conjugate:

$$I_D \propto R_p^2 \cos^2 \Psi_e + R_s^2 \sin^2 \Psi_e + 2 \text{Re} \left( \frac{R_p}{R_s} \right) \sin \Psi_e \cos \Psi_e \sin(2A)$$

$$+ 2 \text{Re} \left( \frac{R_p}{R_s} \right) \sin^2 \Psi_e \cos(2A) \quad \text{A.4}$$

Normalizing A.4 by dividing by the term independent of $A$ yields:

$$I_D \propto 1 + \frac{R_p^2 - \tan^2 \Psi_e}{\frac{R_p}{R_s}} \cdot \cos(2A) + \frac{2 \text{Re} \left( \frac{R_p}{R_s} \right) \tan \Psi_e}{\frac{R_p}{R_s}} \cdot \sin(2A)$$

$$+ \frac{R_p^2 + \tan^2 \Psi_e}{\frac{R_p}{R_s}} + \frac{\left( \frac{R_p}{R_s} \right)^2}{\frac{R_p}{R_s}} + \tan^2 \Psi_e$$

$$\quad \text{A.5}$$

The complex Fresnel reflections are related to the quantities that are to be measured:
\[ \frac{R_p}{R_s} = \tan \Psi \exp(i\Delta) \]  

And then inserting into A.5 yields:

\[ I_D \propto 1 + \frac{\tan^2 \Psi_r - \tan^2 \Psi_e}{\tan^2 \Psi_r + \tan^2 \Psi_e} \cos(2\Delta) + \frac{2 \tan \Psi_r \cos \Delta \tan \Psi_e}{\tan^2 \Psi_r + \tan^2 \Psi_e} \sin(2\Delta) \]  

This equation is in the form of a Fourier series so one can define:

\[ \alpha = \frac{\tan^2 \Psi_r - \tan^2 \Psi_e}{\tan^2 \Psi_r + \tan^2 \Psi_e}; \quad \beta = \frac{2 \tan \Psi_r \cos \Delta \tan \Psi_e}{\tan^2 \Psi_r + \tan^2 \Psi_e} \]  

And since A is a function of time, the detector signal is measured versus time so the Fourier transform can be executed to determine \( \alpha \) and \( \beta \). Thus there is no need to determine A in equation A.7 and eliminates a variable. After finding \( \alpha \) and \( \beta \) and knowing \( \Psi_e \) the azimuthal angle \( \Psi_r \) and phase difference \( \Delta \) can be determined by:

\[ \tan \Psi_r = \sqrt{\frac{1 + \alpha}{1 - \alpha}} |\tan \Psi_e|; \quad \cos \Delta = \frac{\beta}{\sqrt{1 - \alpha^2}} |\tan \Psi_e| \]  

\( \Psi_r \) and \( \Delta \) are both functions of wavelength so many data points are gathered over a range of wavelengths. This yields some material characteristics of the particular sample being studied, but now these results need to be compared to theory to obtain further information about the sample, such as film thickness.

**Bulk Material Theory**

Given that many materials have already been characterized optically there is a great wealth of information to draw upon to apply to theory. For most materials, especially silicon, the complex indices of refraction are already known. Using the known values with theory allows prediction of the behavior of a linearly polarized beam of light.
impinging the surface of a sample. Starting with a single substrate having no other films, important equations will be shown that will carry over to future equations.

The Jones matrix equation approach is the beginning point. For the reflected beam:

$$\begin{pmatrix}
\tilde{E}_p \\
\tilde{E}_s
\end{pmatrix}^{\text{reflected}} =
\begin{bmatrix}
\tilde{r}_p & 0 \\
0 & \tilde{r}_s
\end{bmatrix}
\begin{pmatrix}
\tilde{E}_p \\
\tilde{E}_s
\end{pmatrix}^{\text{incident}}$$

A.10

And similarly for the transmitted beam with $\tilde{t}$ designating the complex Fresnel coefficient instead of $\tilde{r}$. Here again $E$ is the complex electric field and $\tilde{r}$ is the complex Fresnel reflection coefficients where $p$ and $s$ are still the parallel and perpendicular directions. The four equations that arise from this are:

$$\tilde{r}_p \equiv \frac{\tilde{E}_p}{\tilde{E}_p^{\text{incident}}}; \quad \tilde{r}_s \equiv \frac{\tilde{E}_s}{\tilde{E}_s^{\text{incident}}}$$

A.11

$$\tilde{t}_p \equiv \frac{\tilde{E}_p}{\tilde{E}_p^{\text{incident}}}; \quad \tilde{t}_s \equiv \frac{\tilde{E}_s}{\tilde{E}_s^{\text{incident}}}$$

A.12

The electromagnetic waves from the incident, reflected, and transmitted beams with components $p$ and $s$ will each have the general form:

$$\tilde{E}_{p,s}(\tilde{r}_s,\tilde{t}) = \tilde{E}_{0p,s} \exp\left(\frac{i2\pi n_0}{\lambda} q_{\text{incident}} \cdot \tilde{r}\right) \exp(-i\omega t)$$

A.13

Then A.13 can be used for each wave (incident, reflected, and transmitted) and plugged back into A.11 and A12 along with the use of Snells law:

$$n_0 \sin \varphi_0 = n_1 \sin \varphi_1$$

A.14
where \( n \) is the complex index of refraction and the subscript “0” denotes the ambient while “1” denotes the sample and \( \phi \) is the angle measured between the beam and sample normal. Altogether this yields expressions for the Fresnel reflection and transmission coefficients:

\[
\begin{align*}
\tilde{r}_p &= \frac{n_1 \cos \phi_0 - n_0 \cos \phi_1}{n_1 \cos \phi_0 + n_0 \cos \phi_1} \\
\tilde{r}_s &= \frac{n_0 \cos \phi_0 - n_1 \cos \phi_1}{n_0 \cos \phi_0 + n_1 \cos \phi_1} \\
\tilde{t}_p &= \frac{2n_0 \cos \phi_0}{n_1 \cos \phi_0 + n_0 \cos \phi_1} \\
\tilde{t}_s &= \frac{2n_0 \cos \phi_0}{n_0 \cos \phi_0 + n_1 \cos \phi_1}
\end{align*}
\]

Now these equations can be used with A.6 to provide values of \( \Psi_r \) and \( \Delta \) and checked against experimentally measured values.

**Thin Film on a Substrate**

The next step is to add a thin film on top of the substrate and examine the effects. For this, a naming convention will be set for the ambient, film, and substrate using subscripts “0”, “1”, and “2” respectively. The equations shown will be given in the general form but may be applied to both \( p \) and \( s \) directions. To begin, an expression for the electric field is needed that now applies to propagating waves between the top and bottom interfaces. This is given as:

\[ \tilde{E}_{\text{after}} = \tilde{E}_{\text{before}} \exp(-i2\beta) \]
where $\beta$ is the phase thickness of the film that is:

$$\beta = 2\pi d \cos \phi_1 = 2\pi \frac{d}{\lambda} \sqrt{n_1^2 - n_0^2 \sin^2 \phi_0}$$  \hspace{1cm} \text{(A.20)}$$

Here $d$ is the film thickness (unknown) and $\lambda$ is the wavelength. Equation A.19 is applied to every wave reflected, thus a general formula for each wave is:

$$E_n = t_{10} t_{01} \left( r_{10} \right)^{n-2} \left( r_{12} \right)^{n-1} e^{-j(2n-2)\beta}$$  \hspace{1cm} \text{(A.21)}$$

And upon summation over all waves reduces to:

$$E_{\text{total}} = \left[ r_{01} + t_{10} \sum_{n=2}^{\infty} \left( r_{10} \right)^{n-2} \left( r_{12} \right)^{n-1} e^{-j2n\beta} \right] E_{\text{incident}}$$  \hspace{1cm} \text{(A.22)}$$

Using A.15–A.18 will give:

$$r_{01} = -r_{10}$$  \hspace{1cm} \text{(A.23)}$$

$$t_{10} t_{01} = 1 - r_{01}$$  \hspace{1cm} \text{(A.24)}$$

Which are put into A.22 to yield:

$$E_{\text{total}} = \left( \frac{r_{01} + r_{12} e^{-i2\beta}}{1 + r_{01} r_{12} e^{-i2\beta}} \right) E_{\text{incident}}$$  \hspace{1cm} \text{(A.25)}$$

A pseudo-Fresnel reflection coefficient is now defined as:

$$R_p \equiv \frac{E_{\text{total}}}{E_{\text{incident}}} = \frac{r_{01} + r_{12} e^{-i2\beta}}{1 + r_{01} r_{12} e^{-i2\beta}}$$  \hspace{1cm} \text{(A.26)}$$

This equation holds true for p and s directions as long as the values for $r_{mn}$ are calculated for both p and s and used in their respective cases. The $r_{mn}$s are calculated using A.15 and A.16 making sure one keeps track of the reflections from each of the interfaces.
These values are input into A.26 along with the calculated $\beta$ from A.20 to get $\tilde{R}_p$ and $\tilde{R}_s$. Once they have been determined, A.6 will be used to solve $\Psi_r$ and $\Delta$. By making an initial guess for $d$ in A.20, successive iterations to the true value of $d$ will yield values of $\Psi_r$ and $\Delta$ such that they match experimentally determined values of $\Psi_r$ and $\Delta$ found from A.9.

Some final notes need to be mentioned on the analysis of the samples. Many samples contain more than one layer, therefore extensions on the above equations would need to be made. Summing the multiple reflections from the layers can further extend the pseudo-reflection coefficients. The other approach is to assign each layer and substrate its own matrix. The matrices are multiplied together in order and used for the calculations.

The final comment worth stating is the algorithm used to converge on solutions of film thickness. Equations and calculations will not be shown and are left to the reader’s interest. For the Variable Angle Spectroscopic Ellipsometer (VASE) software used in this dissertation, the Levenberg-Marquardt algorithm was used to minimize the Mean Squared Error (MSE) in measurements compared to the model.
Figure A.1. Schematic of an ellipsometer showing light from the polarizer impinging on the surface of a sample and reflecting into the analyzer taken from reference [Hum93].
APPENDIX B
OVERVIEW OF HALL EFFECT AND ITS EQUATIONS

This section displays some of the common equations used in studying Hall effect. The basics will be shown as a foundation to deriving more complex equations used in the analysis. Since there are different cases involved, the general equations are derived first so subsequent equations for the specific cases can be derived directly. In all cases non-uniform, conductive layers are dealt with due to the nature of Gaussian profiles created by ion implantation. Discussions on measurements and equations used by the Hall system will also be related to the theoretical basis. General explanations on Hall effect can also be explored in several sources [Sch98, Li93, Joh70]

Edwin Hall discovered the Hall effect [Hal79] when he observed a deflection in an applied current if the sample is placed in a magnetic field. Hall effect is used extensively in the semiconductor industry to electrically characterize materials as it bears information such as resistivity/conductivity, carrier dose/concentration, and mobilities. The assay begins with applying an electrical current through a material perpendicular to a magnetic field. By doing this, an electric field is generated in a direction orthogonal to both the current and magnetic field. This is depicted in figure B.1. Holes and electrons will be deflected into the negative y direction given that the current and magnetic field are in the positive x and z direction respectively.

The force on holes or electrons is given by:
\[ F = q \left( \xi + v \times B \right) \]  \hspace{1cm} (B.1)

Here \( q \) is the charge of an electron or hole, \( \xi \) is the electric field, \( v \) is the velocity of carriers, and \( B \) is the magnetic field. The force in the \( y \) direction becomes 0 at steady state therefore using the dimensions given and properties of electrically conducting material, the voltage generated in the \( y \) direction, the Hall voltage, \( V_H \), can be determined.

\[ V_H = \frac{BI}{qn_k} \]  \hspace{1cm} (B.2)

where \( I \) is the current in the \( x \) direction and \( n_k \) is the carrier concentration of either electrons or holes. Now a Hall coefficient is defined as:

\[ R_H = \frac{tV_H}{BI} \]  \hspace{1cm} (B.3)

Note that the possibility of having both holes and electrons may exist together so further derivations need to include both carriers. Expressions are now needed for \( V_H \) and \( I \) based on the properties of the materials. The current can be split into \( x \) and \( y \) components then:

\[ I_x = w \left[ q \sum_k \int_0^t n_k(z) \mu_k(z) dz \xi_x + \sum_k \int_0^t q_k \mu_k(z) \left( \frac{J_{xy}}{q} \right) dz B \right] \]  \hspace{1cm} (B.4)

\[ I_y = l \left[ q \sum_k \int_0^t n_k(z) \mu_k(z) dz \xi_y - \sum_k \int_0^t q_k \mu_k(z) \left( \frac{J_{xx}}{q} \right) dz B \right] = 0 \]  \hspace{1cm} (B.5)

The subscript “\( k \)” refers to either electrons or holes. The mobility is \( \mu_k \) of the \( k^{th} \) species, \( \xi \) is the electric field for either the \( x \) or \( y \) direction, and \( J_k \) represents the current flux in either direction. The currents are really just the integration of the current fluxes over the thickness, \( t \), and either the width, \( w \), or length, \( l \), (depending on the direction of current)
of the material. Clearly at steady state $J_{ky}$ and $I_y$ will be zero and thus allows an expression for $J_{kx}$ to be deduced as:

$$J_{kx}(z) = q \sum_k n_k(z) \mu_k(z) \xi_x$$ \hspace{1cm} (B.6)

It is also known that $\xi_y$ is defined as

$$\xi_y = \frac{V_H}{w} \hspace{1cm} (B.7)$$

Using (B.4) – (B.7) and plugging into (B.3) an expression for $R_H$ becomes

$$R_H = \frac{t \sum_k \int_0^l n_k(z) \mu_k^2(z) \, dz}{\left( q \sum_k \int_0^l n_k(z) \mu_k(z) \, dz \right)^2} \hspace{1cm} (B.8)$$

Then using the Hall coefficient the active dose is given as:

$$\varphi = \frac{t}{q R_H} \hspace{1cm} (B.9)$$

Here the active dose is used since implants are performed according to a measured dose so the success of activation can be determined by comparing active dose to implant dose.

Next an expression for the material resistivity is derived starting with the basic equation:

$$\rho = \frac{wt \, V_x}{l \, I_x} \hspace{1cm} (B.10)$$

A similar equation to (B.7) for $\xi_x$ using $l$ instead of $w$ along with (B.4) and (B.10) will give:

$$\rho = \frac{t}{q \sum_k \int_0^l n_k(z) \mu_k(z) \, dz} \hspace{1cm} (B.11)$$

The final material characteristic to define is the Hall mobility:
\[ \mu_H = \frac{|R_H|}{\rho} \] \hspace{1cm} (B.12)

Since B.8 is a general expression for the Hall coefficient, a couple cases need to be considered. Beginning with an implant layer that is well behaved and is isolated from the substrate, equation B.8 reduces to:

\[ R_H = \frac{\int_{0}^{t_1} q n_p(z) \mu_p(z) dz}{q \int_{0}^{t} n_p(z) \mu_p(z) dz} \] \hspace{1cm} (B.13)

Having complete isolation from the substrate is not always possible due to leakage currents, therefore terms for a substrate with homogeneous doping will need to be included so \( R_H \) becomes:

\[ R_H = \frac{\int_{0}^{t_1} q n_p(z) \mu_p(z) dz - q n_n \mu_n t_2}{q^2 \left( \int_{0}^{t} n_p(z) \mu_p(z) dz + n_n \mu_n t_2 \right)^2} \] \hspace{1cm} (B.14)

where \( t_1 \) refers to the junction depth of the implant layer, \( t_2 \) is the thickness of the n-type layer, and \( t \) is the total thickness (=\( t_1 + t_2 \)).

A brief review of the equations used by the Hall system is given to relate the measured values to the theory. Once data has been collected from the Hall system the above theoretical equations can be used along with the known material properties to extract further information about samples. Essentially all the Hall system measures is a voltage potential across two other points within a magnetic field while a know current is applied at two points (four contacts are made to each sample). The first characteristic measured is the resistivity of the active layer from:
\[ \rho = \frac{\pi t F(R_{12,34} + R_{23,41})}{2 \ln 2} \]  

B.15

The subscripts in the resistance, \( R \), define a current forced from contact 1 to contact 2 while a voltage is measured across contact 3 to contact 4 in the first “R” term. Other “R” terms subscripts have similar defining conventions. The factor, \( F \), is a correction factor allocated by:

\[ F = \frac{-2 \ln 2}{\ln a + \ln (1 - a)} \]  

B.16

where

\[ a \text{ satisfies } a^z = 1 - a \]

\[ z = x \text{ if } x \leq 1 \]

\[ \text{or } z = 1/x \text{ if } x > 1 \]  

B.17

and

\[ x = \begin{vmatrix} R_{12,34} \\ R_{23,41} \end{vmatrix} \]  

B.18

Under an applied magnetic field the mobility is measured as:

\[ \mu_{H} = \frac{10^8(\Delta R_{13,24} + \Delta R_{24,31})}{2 \rho B} \]  

B.19

The factor of \( 10^8 \) is included in the equation because \( B \) is usually given in gauss so conversion to metric units standard relies on this factor. With the resistivity and mobility determined, other variables are defined as:

\[ R_{H} = \rho \mu_{H} \]  

B.20

\[ \varphi = \frac{t}{\rho q \mu_{H}} \]  

B.21
where $R_s$ is the sheet resistance of the material. It should be mentioned that $t$ is input by the user of the Hall system and the concentration of the active layer can be determined if the layer has constant active levels throughout the layer. Since the layers involved in this research are created by ion implantation the concentration measured would be an averaged concentration. Altogether these equations will yield know values that are used with equations such as B.9, B.11, B.12, B.13, and B.14.
Figure B.1. Hall measurement of a sample showing how directions of current and magnetic field produce the Hall voltage.
Sub simsintegrater()
Dim A(200) As Double
Dim B(200) As Double
Dim C(200) As Double
Dim area As Double
Dim guess As Double
Dim halldose As Double
Dim adose As Double
Dim adose2 As Double
Dim zmin As Integer
Dim zmin2 As Integer
Dim zmax As Integer
Dim zmax2 As Integer
Dim Z(200) As Integer
Dim y1 As Double
Dim y2 As Double
Dim conc As Double

Dim p As Double
Dim n As Double
Dim mu As Double
Dim mun As Double
Dim t As Double
Dim t2 As Double
Dim RHs As Double
Dim Rs As Double
Dim muH As Double
Dim integral As Double
Dim integral2 As Double
Dim aa As Double
Dim bb As Double
Dim cc As Double
Dim expRHs As Double
Dim t22 As Double
Dim w As Double

guess = Cells(8, 8)
t = Cells(9, 8)
halldose = Cells(10, 8)
expRHs = Cells(11, 8)
n = Cells(12, 8)
mun = Cells(13, 8)

For j = 0 To 199
    A(j) = Cells(j + 4, 3)
    C(j) = Cells(j + 4, 3)
    B(j) = Cells(j + 4, 2)
Next j

area = 0
For l = 0 To 198
    If C(l + 1) > 0 Then
        area = area + 0.5 * (C(l) + C(l + 1)) * (B(l + 1) - B(l)) * 0.0001
    End If
Next l
Cells(7, 10) = area

adose = 0
adose2 = 0
Do Until adose <= halldose And halldose <= adose2 And adose2 > 0 Or adose2 <= halldose And halldose <= adose And adose2 > 0

    adose2 = adose
    adose = 0
    zmax2 = zmax
    zmin2 = zmin

    For i = 0 To 199
        A(i) = C(i)
    Next i

    For i = 0 To 199
        If A(i) > guess Then
            A(i) = guess
        End If
    Next i

    For i = 0 To 198
        If A(i + 1) > 0 Then
            adose = adose + 0.5 * (A(i) + A(i + 1)) * (B(i + 1) - B(i)) * 0.0001
        End If
    Next i
Next i

For i = 0 To 199
    If A(i) < guess Then
        Z(i) = 0
    Else
        Z(i) = i + 1
    End If
Next i

For i = 0 To 198
    If Z(i + 1) > Z(i) Then
        zmax = Z(i) + 1
    End If
Next i

For i = 1 To 199
    If Z(i) = 0 Then
        Z(i) = 200
    End If
Next i

For i = 1 To 198
    If Z(i + 1) < Z(i) Then
        zmin = Z(i + 1) - 2
    End If
Next i

If A(0) = guess Then
    zmin = 0
End If

If adose > halldose Then
    guess = C(zmax)
End If

If adose < halldose Then
    guess = C(zmax - 2)
End If

Loop

If adose > adose2 Then
    zmin = zmin2
End If
If adose2 > adose Then
    zmax = zmax2
End If

y1 = 0
y2 = 0
For i = 0 To 198
    If i < zmin Or i >= zmax And C(i + 1) > 0 Then
        y1 = y1 + 0.5 * (C(i) + C(i + 1)) * (B(i + 1) - B(i)) * 0.0001
    End If
Next i

y2 = (B(zmax - 1) - B(zmin + 1)) * 0.0001

If zmin = 0 Then
    y1 = y1 + 0.5 * C(zmax) * (B(zmax) - B(zmax - 1)) * 0.0001
    y2 = y2 + (B(zmin + 1) - B(zmin)) * 0.0001 + 0.5 * (B(zmax) - B(zmax - 1)) * 0.0001
Else
    y1 = y1 + (0.5 * C(zmin) * (B(zmin + 1) - B(zmin)) * 0.0001) + (0.5 * C(zmax) * (B(zmax) - B(zmax - 1)) * 0.0001
    y2 = y2 + 0.5 * (B(zmin + 1) - B(zmin)) * 0.0001 + 0.5 * (B(zmax) - B(zmax - 1)) * 0.0001
End If

conc = (halldose - y1) / y2

If C(zmin + 1) < conc Then
    zmin = zmin + 1
    GoTo 10
End If

If zmin = 0 And C(0) < conc Then
    y1 = y1 + (0.5 * C(zmin) * (B(zmin + 1) - B(zmin)) * 0.0001)
    y2 = y2 - 0.5 * (B(zmin + 1) - B(zmin)) * 0.0001
    conc = (halldose - y1) / y2
End If

adose = 0
For i = 0 To 199
    A(i) = Cells(i + 4, 3)
    If A(i) > conc Then
        A(i) = conc
    End If
Next i

For i = 0 To 198
  If A(i + 1) > 0 Then
    adose = adose + 0.5 * (A(i) + A(i + 1)) * (B(i + 1) - B(i)) * 0.0001
  End If
Next i

For i = 0 To 198
  If B(i + 1) <= t Then
    p = (A(i) + A(i + 1)) / 2
    mu = 44.9 * Exp(-9.23E+16 / p) + 470.5 / (1 + (p / 2.23E+17) ^ 0.719) - 29 / (1 + (6.1E+20 / p) ^ 2)
    integral = integral + p * mu * (B(i + 1) - B(i)) * 0.0001
    integral2 = integral2 + p * (mu ^ 2) * (B(i + 1) - B(i)) * 0.0001
  Else
    GoTo 20
  End If
Next i

20

RHS = integral2 / (1.602E-19 * integral ^ 2)
Rs = 1 / (1.602E-19 * integral)
muH = integral2 / integral
aa = expRHs * (1.602E-19) * (n ^ 2) * (mun ^ 2)
bb = 2 * expRHs * (1.602E-19) * n * mun * integral + n * (mun ^ 2)
cc = expRHs * (1.602E-19) * (integral ^ 2) - integral2
w = bb ^ 2 - 4 * aa * cc

If w < 0 Then
  t2 = 0
  t22 = 0
Else
  t2 = (-bb + ((bb ^ 2 - 4 * aa * cc) ^ 0.5)) / (2 * aa) * 10000
  t22 = (-bb - ((bb ^ 2 - 4 * aa * cc) ^ 0.5)) / (2 * aa) * 10000
End If

Cells(8, 10) = adose
Cells(10, 10) = conc
Cells(11, 10) = Rs
Cells(12, 10) = RHS
Cells(13, 10) = muH
Cells(14, 10) = t2
Cells(15, 10) = t22

For j = 0 To 199
    If A(j) > 0 Then
        Cells(j + 4, 5) = A(j)
    End If
Next j

End Sub
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<td>Thickness (µm): 0.0093</td>
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<td>n-type RHs (cm²/C): -24245.71</td>
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Figure C.1. Sample spreadsheet showing input parameters entered by the user and theoretical values calculated by the program.
LIST OF REFERENCES


<table>
<thead>
<tr>
<th>Citation</th>
<th>Reference</th>
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</table>


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BIOGRAPHICAL SKETCH

Chad Darrell Lindfors was born to Reed and Judy Lindfors on June 8, 1974, in Fridley, Minnesota. Growing up in Blaine, Minnesota, he decided to stay and attend school at the University of Minnesota. In 1997 he received his B. S. degree in chemical engineering while keeping a strong interest in microelectronics. While in school he started working at 3M Company where he remained until 1998. In August of 1998 he joined the Materials Science and Engineering Department at the University of Florida in pursuit of his Ph.D. degree. He received his M. S. degree in August of 2001 and interned at Motorola Inc. the same summer. His current research interest is focused on the study of dopant activation and material characteristics of ion-implanted silicon materials using solid phase epitaxy regrowth techniques. Some days the best place to find him is on top of a mountain.