

ELECTROMAGNETIC FAST-FIRING FOR THE RAPID THERMAL PROCESSING OF SILICON

By

Keith Thompson

A dissertation submitted in partial fulfillment

of the requirements for the degree of

Doctor of Philosophy

(Electrical Engineering)

at the

UNIVERSITY OF WISCONSIN, MADISON

2003

UMI Number: 3089590

UMI[®]

UMI Microform 3089590

Copyright 2003 by ProQuest Information and Learning Company.

All rights reserved. This microform edition is protected against
unauthorized copying under Title 17, United States Code.

ProQuest Information and Learning Company
300 North Zeeb Road
P.O. Box 1346
Ann Arbor, MI 48106-1346

Petitioner Micron Ex-1059, 0002

A dissertation entitled

ELECTROMAGNETIC FAST-FIRING FOR THE
RAPID THERMAL PROCESSING OF SILICON

submitted to the Graduate School of the
University of Wisconsin-Madison
in partial fulfillment of the requirements for the
degree of Doctor of Philosophy

by

KEITH JOSEPH THOMPSON

Date of Final Oral Examination: FEBRUARY 25, 2003

Month & Year Degree to be awarded: December

May 2003

August

Approval Signatures of Dissertation Committee

Reid T. Cooper

John D. Puskas

Thomas J. Plesch

Susan C. Ogden

Brian Brand

Robert D. Johnson

Signature, Dean of Graduate School

Martin C. Caldwell, Jr.

ABSTRACT

The application of high temperature, short-term thermal processing, RTP, techniques is critically important in silicon processing technology. The creation of high dose ($>10^{20}/\text{cm}^3$), low resistivity (<600 ohm/square), ultra-shallow junctions (< 25 nm), requires RTP techniques that restrict diffusion and activate a significant fraction of the implanted dopant species. A new technology, electromagnetic induction heating (EMIH), achieves rapid heating by directly coupling electromagnetic radiation into the silicon wafer. Heating rates of $125^\circ\text{C}/\text{sec}$ to temperatures in excess of 1050°C were consistently achieved for 75 mm and 100 mm diameter wafers at input powers of 1000 and 1300 watts respectively. These ramp rates are suitable for ultra shallow junction formation, and junctions shallower than 25 nm with sheet resistances lower than 600 ohms/square have been achieved.

This work describes the application of electromagnetic heating using radiation in the radio, 13.56 MHz, and microwave, 2450 MHz, frequency regimes. Maxwell's equations are used to detail the theory behind EMIH, to describe the heating phenomena, to discuss efficient field-wafer coupling, and to analyze the material dependence on heating. The experimental systems constructed to implement this heating technique, along with heating results as a function of input power are also provided.

Electromagnetic heating was simulated through finite difference time domain calculations, and it was determined that the transverse electric fields drive the heating the process. These simulations predicted a center cool, radial non-uniformity when heating is performed in the TM₀₁₁ mode, an hour glass shaped temperature non-uniformity when heating is performed in the TM₁₁₁ mode, and a center hot, radial non-uniformity when

heating is performed in the TM111 mode while the wafer is rotated at 60 RPM. These simulations match the experimental results, and a 3.6% temperature non-uniformity was achieved during heating in the TM111 mode when the wafer was rotated at 60 RPM

The application of EMIH was illustrated by creating ultra-shallow junctions and by studying the transport and activation of boron in silicon under a variety of conditions. The ultra-shallow junctions created satisfy the 90 nm CMOS technology node, as defined by SEMATECH, and are comparable to those achieved with state-of-the art lamp-based anneal technologies. The transport of boron was directly linked to the concentration of free boron interstitials, B_i , and silicon vacancies, V_{Si} . Si interstitials, Si_i , and oxygen atoms were found to annihilate V_{Si} , limiting activation and increasing boron transport. The oxidation of the silicon surface by ambient oxygen also consumed boron dose and led to an increased out-diffusion of F from the Si wafer. This detracted from the quality of the junctions in two ways. First, the consumption of dose increased the final sheet resistance. Second, fluorine atoms were found to bind up Si_i and B_i in Si-F and B-F bonded complexes. This limited boron transport and prevented the Si_i from annihilating V_{Si} , thereby improving the statistical likelihood of boron activation. Finally, EMIH at 2.45 GHz appears to have an optimal anneal temperature that is 100°C lower than that of comparable lamp-based anneal systems. This either indicates the existence of a microwave induced ponderomotive force or illustrates the role of optical illumination in the diffusion and activation of boron in silicon.

Finally, EMIH was applied to silicon-silicon and silicon-silicon dioxide bonding to illustrate that the rapid bonding of silicon wafers is possible. Full strength, void free silicon-silicon bonds were achieved in 5 minutes at 1000°C.

ACKNOWLEDGMENTS

I would like to express my deepest gratitude to my co-advisors, John H. Booske and Reid F. Cooper, who acted as a constant support for my work at the UW. They provided a wealth of innovative ideas and were the source of numerous lengthy discussions, and at least as many lengthier tangents. It was through their ever-challenging nature that this document has evolved from a primitive reporting of results to an involved explanation of the basic physical phenomena surrounding my field of work. They were also gracious enough to allow me the opportunity to travel, to present, and to defend my work in front of a mixed academic-industrial community. The feedback I received at these conferences was instrumental in advancing my own investigations.

I would further like to thank Dan Downey, Sue Felch, Edwin Arevalo, Reuel Liebert, Tony Renau, and the entire research staff at Varian Semiconductor Equipment Associates. Dan played an especially critical role in helping bring my work to the cutting edge of short-time, high-temperature processing. He provided an innumerable number of shallow implanted samples and post-anneal analysis, and he was an incredible resource for discussions and ideas. These ideas are referenced within this document and opened up numerous avenues of investigation that I had not initially considered. Dan also afforded me the wonderful opportunity to spend 4 months interning at Varian. Working with Dan was an incredible experience for too many reasons to list, not the least of which was the fact that the four months I spent in Gloucester, Mass were summer months☺!

Additional thanks goes out to Yogesh Gianchandani who provided consistent support through my time at UW. The rapid microwave bonding and microwave MEMS work, which played a larger educational role than discussed in this manuscript, can be largely attributed to Yogesh. I would also like to thank Susan Hagness who introduced me to the world of electromagnetic simulations. She literally taught me everything that I now know about simulating EM fields, specifically Finite Difference Time Domain, FDTD, simulations. The FDTD simulation work is largely due to her patience and guidance. Finally, I would like to thank all of those people who helped me throughout the last three years whom I've failed to individually acknowledge here. Any body or research worth reporting cannot be accomplished without the help of dozens of associates ranging from technicians, who typically do the lion's share of the unacknowledged work, to friendly associates and colleagues, who often act as a sounding board for ideas and to provide moral support. Many thanks to all involved.

Keith J. Thompson

TABLE OF CONTENTS

Abstract	i
Acknowledgments	iii
Chapter 1: Background	1
1.1 Scaling	2
1.2 Damage Driven Diffusion	6
1.3 Motivation for Microwave RTP	9
1.4 Other Applications of EMIH	11
Chapter 2: Theory of Electromagnetic Induction Heating	12
2.1 Maxwell's Laws	13
2.2 Field to Wafer Coupling	14
2.3 Material Dependence	16
Chapter 3: Experimental Apparatus	21
3.1 Microwave Heating	21
3.2 RF Heating	23
3.3 Temperature Measurement	24
Chapter 4: Heating Results	25
4.1 RF Heating	25
4.2 Microwave Heating	29
4.3 Microwave Simulations	33
4.3.1 Quality Factor Calculation	35
4.3.2 Uniformity	36

Chapter 5: Ultra-Shallow Junction Results	50
5.1 Principles of Diffusion and Activation	51
5.2 Concentration Enhanced Diffusion	58
5.3 Maximizing Activation, Minimizing Diffusion	60
5.4 Effect of Environmental Oxygen on Boron Diffusion and Activation	62
5.5 Effect of Implanted Fluorine Boron Diffusion and Activation of Boron	78
5.6 Effect of Illumination on Boron Diffusion and Activation	85
5.7 Ultra-Shallow Junction Comparison	90
Chapter 6: Bonding Results	93
6.1 Microwave Bonding	93
6.2 RF Bonding	96
6.3 Surface Preparation for Rapid Bonding	97
6.4 Silicon on Insulator Bonding	99
6.5 Multi-Wafer Bonding	100
Chapter 7: Conclusion	101
Appendix	106
List of Figures	109
References	115

CHAPTER 1: BACKGROUND

Digital technology relies upon the concept of storing and retrieving bits of information in a simple "yes" or "no", 0/1, format. The basic switch, the ability to differentiate between an "on" and "off" state, represents the foundation of digital logic implementation.

In semiconductor microelectronics, the most efficient way to obtain this switch is through Complementary Metal Oxide Semiconductor, CMOS, devices. These devices, Fig. 1-1, work by a basic counter-doping scheme. The source/drain junction regions are chemically altered, through a process called doping [1], so that

the electrical polarity of their mobile charge carriers directly opposes that of the well or "channel" region located underneath the gate dielectric,

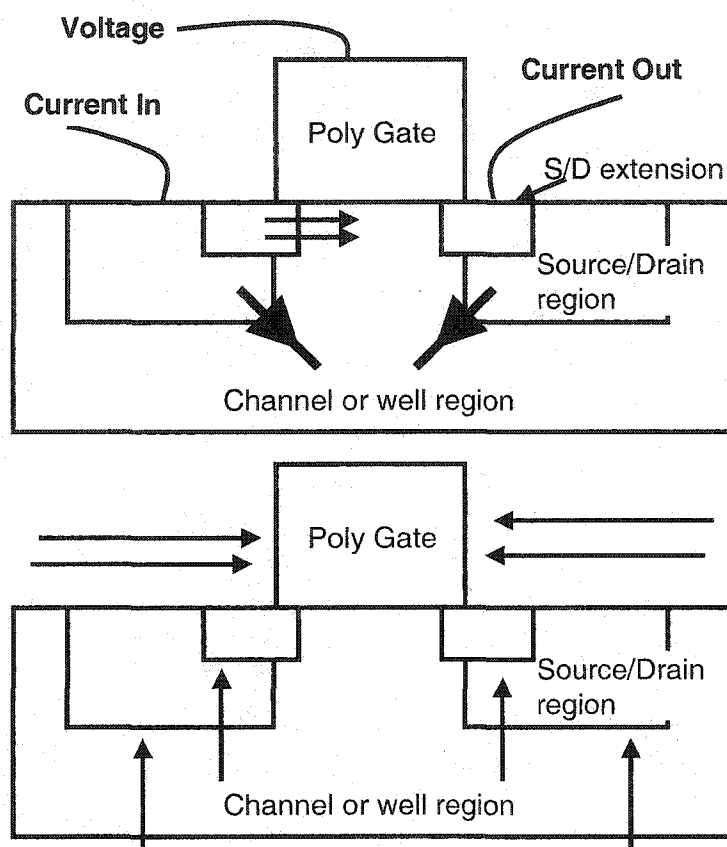


Fig. 1-1. Top: A basic MOS device structure. The diode symbols show where the diode effect electrically isolates the doped regions, and the solid arrows indicate where the current and voltage is applied/drawn.

Bottom: A MOS device with arrows showing the scaling directions for future devices.

which separates the source/drain regions. The opposing polarity leads to a potential barrier, or diode effect, at the interface of these doped regions. The symmetry of these diodes allows current to flow into or out of the well region, but not both. Applying a voltage to the gate, whose polarity matches that of the majority carrier type in the region just under the gate dielectric, depletes this region of its mobile charge carriers. When this voltage is strong enough, mobile carriers of the opposite polarity are attracted into the gate region. This effectively inverts the polarity of the channel region, eliminating the potential barriers or diode effect, and allowing current to flow between the source and the drain. This applied gate voltage is the switch that flips the transistor between the "off" and "on" states.

1.1 Scaling

Reducing the distance between the source and drain regions, more commonly known as the effective gate length, increases both the switching speed of the transistor and the number of transistors that can be built on a wafer [2,3]. This simultaneously enhances the performance of the circuit fabricated while decreasing the per die manufacturing cost. Because of the three dimensional nature of these devices, scaling in one direction requires scaling in all directions. In other words, as transistor dimensions scale horizontally, junction depths must scale vertically. For example, by 2003, according to the International Technology Roadmap on Semiconductors (ITRS) [4], the 100 nm node – with physical gate lengths of 45 nm, maximum junction depths of 31 nm, and minimum active dopant concentrations of $10^{20}/\text{cm}^3$ – must enter production. By 2013 the physical gate lengths will shrink to 18 nm, requiring junction depths of 9 nm or less.

The benefits of scaling do not come without significant drawbacks. The source/drain regions act as the contacts between the device and interconnect lines. Because power dissipation at these contacts follows the well

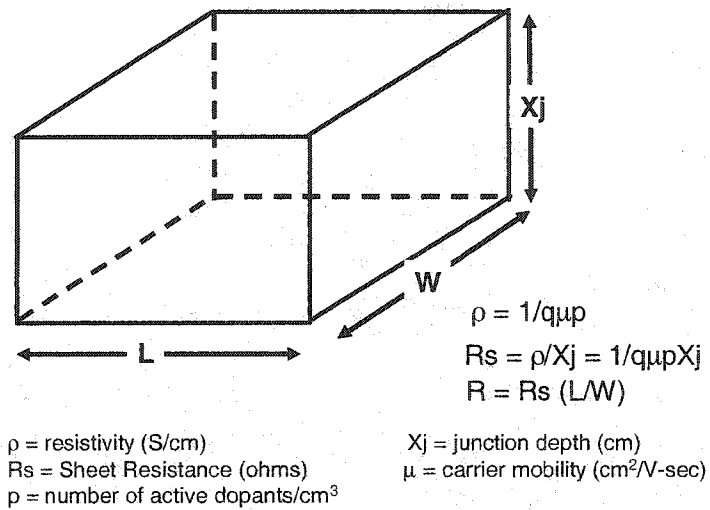


Fig. 1-2. A shallow junction modeled as a resistor

known $P = I^2 R$ formula, increasing the resistance of these contacts directly impacts the power dissipated during the switching of the transistor. Modeling the junction region as a resistor, Fig. 1-2, illustrates that reducing the depth of the source/drain regions, without changing the material properties of the silicon, increases the junction resistance. The basic equation for resistance is:

$$R = \frac{\rho}{X_j} \left(\frac{l}{w} \right) \quad (1-1)$$

where ρ is the material resistivity, w is the width, l is the length, and X_j is the depth of the transistor, as seen in Fig. 1-2. This equation is written more conveniently as:

$$R = R_s \left(\frac{l}{w} \right) \quad (1-2)$$

$$R_s = \frac{\rho}{X_j} \quad (1-3)$$

R_s is the sheet resistance of the material and is a more useful reference since CMOS scaling typically assumes that the (l/w) ratio remains unchanged. The final equation, eqn. 1-3, clearly illustrates that for a given resistivity the sheet resistance depends inversely on the

junction depth. The classic R_s - X_j trade-off curve, Fig. 1-3, illustrates this relationship. The boxes in this figure indicate the scaling requirements for each CMOS technology node, and the dashed line

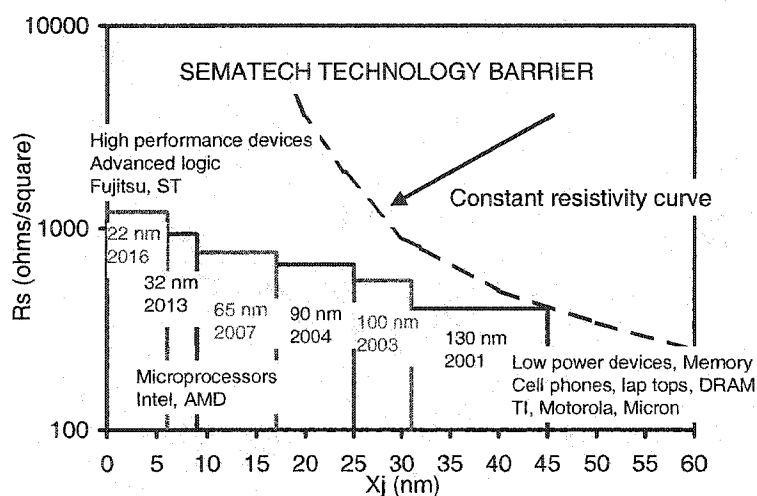


Fig. 1-3. The classic sheet resistance (R_s) – junction depth (X_j) tradeoff

illustrates the effect of reducing junction depth without changing the material properties of the silicon.

The cost of superior device performance (smaller transistors) is a higher level of power dissipation. As the devices scale in volume the amount of current that flows through each device remains constant. The current is the number of electrons that pass through a cross-section of silicon each second. In a junction, the cross-section is composed of the width of the transistor, W , and the depth of the transistor, X_j . Although the current remains constant, the current density increases as the cross-sectional area decreases. Since W and X_j both decrease with future technology nodes, as the number of devices built in a given region increases, the total current increases. For example, cutting critical dimensions in half quadruples the number of devices fabricated. The current density quadruples as does the total current dissipated. Further exacerbating this issue is the increase in resistance that occurs as the junction depth scales. Because resistance scales inversely with junction depth (see eqn. 3 and Fig. 1-3), and because power is equal to the current squared times the junction resistance,

reducing the junction depth without reducing the silicon resistivity results in a monotonic increase in the amount of dissipated power per device.

While some high performance devices can tolerate higher resistive losses in exchange for rapid switching (i.e. advanced logic devices composed of relatively few transistors and with access to a fixed power supply), low power devices operate at more moderate switching speeds to conserve energy (i.e. simple portable electronic devices such as cell phones and pagers). The most advanced devices, however, require both low power and high performance (i.e. microprocessors, especially for laptops and high end servers) and cannot tolerate this trade-off.

There are other requirements to transistor scaling. Proper electrical isolation between the source and drain during the off state, requires at least 2 orders of magnitude difference in active dopant concentration between the source/drain and channel regions. Rapid switching implies low capacitance, which entails minimizing the gate-source/drain and source/drain-bulk capacitance. At least 6 orders of magnitude should separate the on/off current when the maximum drain-to-source bias is applied. Finally, encroachment of the source/drain dopants into the channel region causes short channel effects (roll-off of the threshold voltage [5,6]), which detrimentally lower the threshold voltage used to invert the channel region. All of these require a highly activated dopant concentration ($\sim 10^{20}/\text{cm}^3$), minimal residual damage, and excellent lateral abruptness of the dopant profile (minimal diffusion of the dopants from the source/drain regions into the channel region).

The only way to scale transistors without degrading the quality of device performance or completely redesigning the device structure is to reduce the resistivity of the silicon in the junction regions. Because the resistivity of silicon is inversely proportional to the mobile

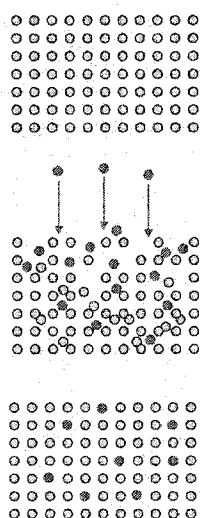
carrier concentration, Fig. 1-2, increasing the volumetric concentration of active dopants is one way to accomplish the necessary reduction in resistivity. This reduction in resistivity offsets the junction depth reduction and maintains the same sheet resistance. This causes the dashed line in Fig. 1-3 to move straight west (left), and improves device performance without incurring higher power losses. Furthermore, a higher dopant concentration reduces short channel effects [7], improves isolation between the source/drain and channel regions, and lowers the background current of the "off" state. Overall, this creates lower power, higher speed devices.

1.2 Damage Driven Diffusion

Achieving these very high levels of dopant concentration in this very shallow region requires the use of low energy, high current ion implantation. Unfortunately, as illustrated in Fig. 1-4, the implantation process severely damages the silicon surface and leaves most of the boron sitting in interstitial sites. Eliminating the damage and moving the implanted dopants into substitutional sites, or "activating" them is a pre-requisite to efficient device operation. This requires a high temperature ($>900^{\circ}\text{C}$) anneal process, which drives the implanted dopants into the wafer, increasing the junction depth, and diluting the volumetric dopant concentration.

Both n-type (As, P) and p-type (B) dopants diffuse by an interstitial mechanism. As a result, mobility is inversely dependent on ionic radii. Smaller atoms have a greater mobility than larger atoms. The small size of ionic B^{3+} , 0.027 nm [1], inside a silicon lattice with an atomic spacing of ~ 0.5 nm, compared to n-type dopants, As (0.046 nm) and P (0.038 nm) [1], implies that n-type dopant diffusion can be expected to be much slower than that of

boron. The mobility of electrons (generated by n-type dopants) is approximately 4 times that of holes (generated by p-type dopants). As a result, n-type layers typically have a lower resistance and shallower junction depth than p-type layers. Consequently,



Perfect Silicon Crystal

All the silicon atoms are aligned perfectly with respect to each other

- Silicon Atom
- Boron Atom

Ion Implantation Doping

The bombarding ions damage the crystal, knocking the silicon atoms from their "designated" sites.

This creates vacancies, interstitials, and extended defects

Anneal Process

1. Repair the damage so that the atoms are once again perfectly aligned.
2. Move the implanted boron atoms onto silicon sites so that they can become electrically *active* (able to donate/capture conduction electrons).

Fig. 1-4. Ion implantation process.

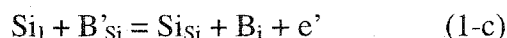
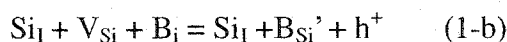
this thesis will focus specifically on the diffusion and activation of boron.

The damage introduced during the implantation process further enhances the diffusion of implanted dopants. Numerous theories have been proposed to explain the anomalous movement of boron (i.e. ultra-fast diffusion) in the first few seconds of the anneal process [8 – 13]. While a great deal of thought and research has gone into these theories, the precise mechanisms for this anomalous diffusion have not been fully confirmed by experiments. Consequently, the following discussion will focus only on the point-defect interactions involved with the diffusion-activation of boron in silicon, leaving further pursuit of this topic to the reader.

It is generally accepted that the damage creates a super-saturation of silicon interstitials, Si_i , which cluster together into extended defects along the $\{311\}$ direction [8 – 13]. The concentration of silicon interstitials is related to both the implant energy and the boron dose. During the implant process, each implanted boron atom has a momentum related to the implant energy used to accelerate it. Higher implant energies result in higher momentum.

The high-momentum boron atoms interact with dozens of silicon atoms, knocking them from their tetrahedral lattice sites and creating Si_I . This process also decelerates the boron atom, causing it to come to rest in an interstitial site, B_I , in the silicon lattice. A higher implant dose (more atoms accelerated into the wafer) leads to a higher concentration of Si_I , and every Si_I created gives rise to a corresponding silicon vacancy, V_{Si} . Because higher energy implants spread the damage deeper into the wafer (on average it takes more collisions to stop the implanted atom), they do not increase the volumetric concentration of Si_I . The depth at which the {311} clusters form, however, is primarily energy dependent. In particular, it increases with implant energy as the damage extends deeper into the wafer.

As the {311} defects dissolve, the released Si_I may recombine with the V_{Si} . The competing reactions are:



The prime, ', denotes a negative ionic charge associated with the active boron. The energy required for a Si_I to fall back into a silicon vacancy, V_{Si} , (rxn. a) is slightly less than 1 eV, and for a boron atom to sit in a V_{Si} , (rxn. b) it is approximately 4.9 eV. Reaction (c) represents the de-activation of a boron from a silicon site by a Si_I . This process is termed "kick-out" and it has, by subtraction of (b) from (c), an activation energy around -3.9 eV. Although deactivation is a strongly exothermic reaction, a large concentration of activated boron atoms is possible by taking advantage of the statistical – and transient – nature of point defect reactions. This entails maximizing the number of implanted boron atoms and silicon vacancies and minimizing the number of silicon interstitials. Unfortunately, the creation of

each silicon vacancy results in the creation of exactly one silicon interstitial. This makes this process – i.e. the creation of regions of super-saturated active dopants – thermodynamically challenging, if not impossible. Two factors make this process kinetically possible. First, the B interstitials are more mobile than the Si interstitials. This means that the B interstitials are able to sample a larger volume of the silicon lattice, in which it can find and occupy a silicon vacancy, than the Si interstitials. Second, the Si_i are trapped in long metastable {311} clusters after the implant, providing a window of time within which strong activation can occur.

1.3 Motivation for microwave RTP

The metastable Si {311} defect clusters begin to dissolve, at a rate fast enough to be detected by measurements, around 700°C. Traditional interstitial (non-damage related) diffusion of boron begins, at a detectable rate, at about 800°C [10,11]. The dissolution of the {311} clusters also gives rise to transient enhanced diffusion (TED), which proceeds orders of magnitude faster than traditional diffusion and continues until the Si interstitial clusters are eliminated. The boron activation process, on the other hand, does not begin in earnest until at least 900°C [20,21]. Fig. 1-5 illustrates, in a comparative fashion, the kinetic mechanisms that dominate in each temperature regime. It is clear from Fig. 1-5 that any time spent in the 700°C – 900°C range enhances deactivation and diffusion without contributing to activation. Taking full advantage of the large concentration of V_{Si} and B_i , without releasing the large quantity of temporarily trapped Si_i requires annealing the wafer for a limited time at a temperature in excess of 1000°C without allowing the wafer temperature to reside in the 700°C – 900°C temperature regime. Minimizing the time spent at 1000°C limits both "kick-

out" and traditional diffusion processes.

The inherent challenge to this thermal strategy, and motivation for rapid thermal processing (RTP), is that to get above 900°C, the wafer temperature must ramp through the lower temperatures at which these parasitic diffusion

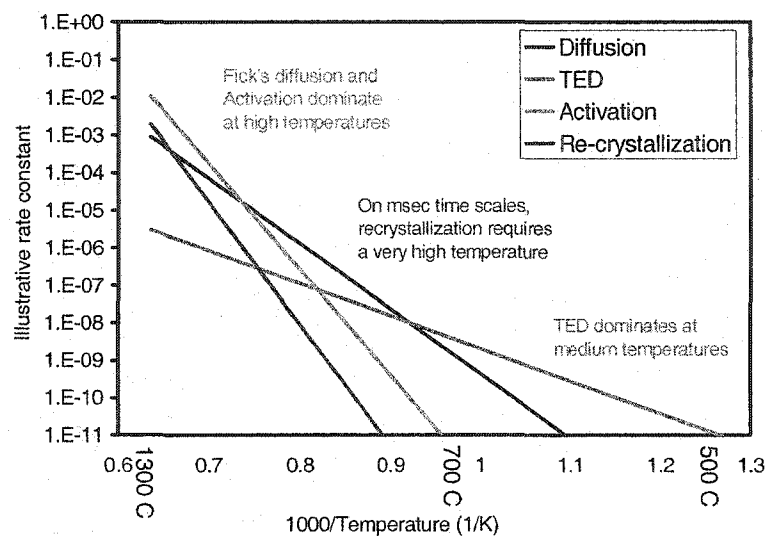


Fig. 1-5. Arrhenius plot illustrating the motivation for rapid thermal annealing of the implanted dopants.

reactions occur. Minimizing this diffusion requires that the wafer is (1) rapidly ramped to the target anneal temperature, typically between 950°C – 1050°C; (2) soaked for the desired amount of time; and (3) cooled back below 500°C as rapidly as possible.

Electromagnetic magnetic induction heating (EMIH), provides ramp rates greater than 125°C/sec, thereby minimizing the amount of time spent in the "parasitic" regime during the ramp up step. Furthermore, the possibility of millisecond EMIH heating, utilizing very high power microwave sources, opens the door for even greater improvements in activation with no net-diffusion of the implanted dopants. Finally, because EMIH is inherently different from lamp-based RTP, it provides an area of basic research. EMIH lacks high intensity optical illumination, heats volumetrically (lamp-based systems heat through surface absorption), and high frequency electric fields are present during the heating process. All of these issues make EMIH an area of study worth pursuing in the hopes of discerning specific phenomena related to the diffusion and activation of boron in silicon that are unique to

electromagnetic RTP and which may lead to an improved process window for ultra shallow junction formation.

1.4 Other applications of EMIH

Besides ultra-shallow junction formation, EMIH also serves as a heating technique for other silicon processes. Specifically, the volumetric heating provide by EMIH makes it very suitable for silicon-silicon and silicon-silicon dioxide wafer bonding. Because the temperatures can be rapidly ramped to the bonding temperature of interest, the bonds can be set in a matter of minutes. This compares favorably to the hour-long anneals necessary using more conventional heating techniques.

CHAPTER 2: THEORY OF ELECTROMAGNETIC INDUCTION HEATING

Creating a scenario in which a material absorbs a majority of incident electromagnetic radiation, as opposed to reflection or transmission, is quite challenging. It had, in fact, been concluded that electromagnetic radiation in the microwave and RF spectral regimes would not couple efficiently (i.e. to produce rapid heating) into silicon without the use of a susceptor [12,14]. Of course, it is known that ohmic heating of silicon boules can be effective through the induction of AC currents at low frequencies, e.g. less than 1 MHz [15]. This heating mechanism is used to remove impurities from the silicon boule during the early stages of high-quality silicon wafer production. As a result, the heating rates are slow and occur over a long period of time as the boule is passed slowly (few millimeters/minute) through the induction coils. Until now, it had been concluded that at frequencies less than 100 GHz the rapid, high temperature heating of silicon was not possible because the ohmic loss tangent associated with silicon was too weak [16]. Because of this, the direct induction heating of silicon wafers was thought impractical, and it was felt necessary to use a susceptor or to add a highly conductive eutectic glue layer (e.g. gold) at the interface to enhance the ohmic response [14]. The experiments performed over the last two years and incorporated into this work have shown conclusively that this is not true. The following describes the fundamental engineering necessary to couple EM energy into a material.

2.1 Maxwell's Laws

Electromagnetic induction heating is a fundamental extension of Maxwell's equations. A magnetic field oscillating transverse to a conducting (or semiconducting) surface induces a similarly oscillating electric flux along that surface. This electric flux, in turn, drives free electrons, in the form of a surface current aligned transverse to the incident magnetic field, through the material. These energetic electrons, upon collisions with the lattice, ohmically transfer their energy to the material, thereby heating it. EMIH differs from previous attempts to heat silicon with microwave radiation [12,14] in that it focuses on maximizing the magnetic field transverse to the wafer surface. Exactly why this technique efficiently heats silicon to temperatures in excess of 1000°C whereas other techniques do not requires a more in-depth study of how electromagnetic fields interact with materials.

Power absorption from the electric field component of a harmonically oscillating electromagnetic field existing within a non-magnetic material can be described by [16]:

$$P_{abs} = \left[\frac{1}{2} \omega_o \epsilon_o \epsilon' \tan \delta \right] |E|^2 \quad \text{Watts/m}^3 \quad (2-1)$$

where ϵ_o is the permittivity of free space, ϵ' is the real component of the dielectric response, ω_o is the frequency of operation, $|E|$ is the magnitude of the electric field inside the material, and $\tan \delta$ is the loss tangent of the material. This loss tangent can be separated into dipole relaxation and conduction current (ohmic) components:

$$\tan \delta = \tan \delta_a + \tan \delta_s = \frac{\epsilon''}{\epsilon' \epsilon_o} + \frac{\sigma}{\omega_o \epsilon' \epsilon_o} \quad (2-2)$$

where σ represents the material conductivity and ϵ'' the imaginary part of the dielectric permittivity associated with polarization relaxation. Eqn. (2-2) in combination with eqn. (2-1) indicates that the dissipation mechanism depends on the four parameters: σ , ϵ' , ϵ'' , and ω_0 . For conventional conductors with large σ and $\epsilon' \sim 1$, ohmic dissipation dominates. For conventional insulators with small σ , dipole relaxation characterized by ϵ'' dominates. For semiconductors the dissipation mechanism varies between dipole relaxation and ohmic conduction losses, depending on the frequency and conductivity. At low conductivity (e.g. low temperature and low impurity doping) and high frequency (e.g. 100 GHz), dipole relaxation, i.e. dielectric losses, dominate. At moderate or high conductivity (high temperature or high doping) or lower frequency (<100 GHz), ohmic conduction losses will usually dominate. Finally, for a medium in which ohmic losses dominate dielectric losses, $\tan \delta$ can be replaced by $\sigma/\omega_0\epsilon_0\epsilon'$ in eqn. (2-2) to obtain the more familiar expression for ohmic power absorption:

$$P_{abs} = \frac{1}{2} \sigma E^2 \quad (2-3)$$

2.2 Field-Wafer Coupling

In addition to having a large $\tan \delta$, efficient electromagnetic heating also requires coupling the electromagnetic fields into the medium heated. From this perspective, efficient coupling occurs when the load (the silicon wafer to be heated) is "well matched" to the electromagnetic fields. Because the properties of the load are fixed when the heating begins, the fields must be tailored to closely match the load. If not, significant wave reflection at the

wafer surface results in poor coupling to the wafer. Equation 2-4 describes the complex impedance of electromagnetic plane waves in a non-magnetic medium [16]:

$$\eta = E_c / H_c = \sqrt{\frac{\mu_o}{\epsilon \left(1 + j\sigma / \omega \epsilon \right)}} \quad (2-4)$$

where η is the wave impedance in the medium, μ_o is the permeability of free space, E_c and H_c are the complex electric and magnetic field phasors respectively, and the other variables are as previously defined. With eqn., 2-4, once can calculate the proper electric to magnetic field phasor ratio, at the surface of the heated material. A dielectric, for example, has a conductivity that approaches zero, so $\eta \sim 377$ ohms (for $\epsilon' = 1$), indicating an E_c/H_c ratio that should be quite large for effective field coupling. For conductors, however, the impedance reduces to:

$$|\eta| = \left| \frac{E_c}{H_c} \right| = \sqrt{\mu_o \omega / \sigma} \quad (2-5)$$

From this relation one can determine, for the range of conductivity observed in silicon, that efficient plane wave field penetration (for which $E_c/H_c \sim 377$ ohms), requires rather high frequencies, i.e. $f > 30$ GHz. This agrees well with the observed limitations discussed at the beginning of this section. At microwave (and lower) frequencies, the impedance is significantly lower. Because of this, efficient coupling requires a magnetic field amplitude that is significantly larger than the electric field amplitude at the wafer surface. Plane wave radiation incident from free space can not provide this ratio.

Two separate techniques for accomplishing this small electric to magnetic field ratio, for efficient EMIH, are described. The first, a resonant cavity, involves placing the wafer at the

bottom of the cavity where the electric field is forced to be small (almost but not quite zero) and the magnetic field is at a maximum. The second involves placing the wafer in the extreme near field of a spiral antenna where the field configuration is known to give large magnetic and small electric fields. More is said about this in the experimental apparatus section.

2.3 Material dependence

With EMIH, the induced current flow is directly dependent on the free carrier density. For rapid high temperature heating, this implies that the free electron density should either increase, or at least not decrease appreciably, with temperature. If the electron density were to decrease with temperature, the current would decrease as well, self-limiting the maximum temperature. On the other hand, if the carrier density were to increase with temperature the heating would be accelerated through thermal runaway, especially if the mobility of the charged species does not change appreciably with temperature. The latter situation is optimal as it results in a very rapid heating of the material.

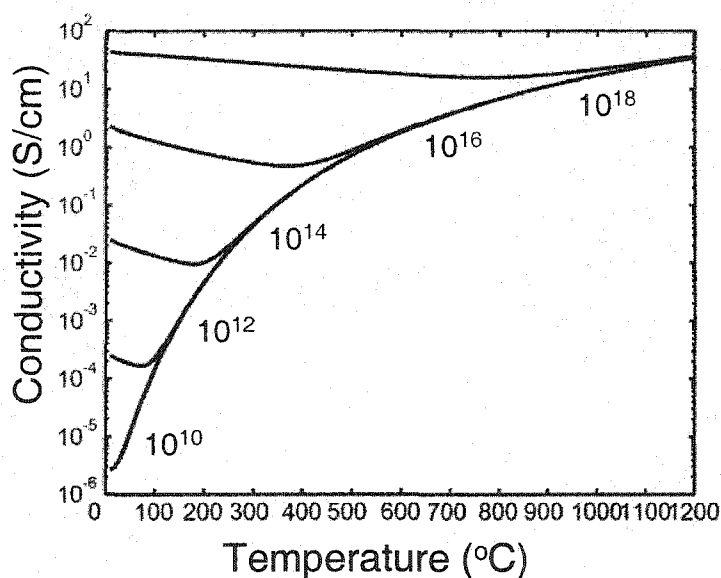


Fig. 2-1. Conductivity of silicon as a function of temperature for several different doping levels.

The material properties of silicon make it a suitable candidate for rapid EMIH heating. The conductivity is large enough for significant levels of current to flow, and most importantly, the free carrier density is extremely sensitive to temperature. As the thermal energy of silicon increases, the number of electrons elevated to the conduction band grows exponentially, increasing the number of carriers available to couple to the electromagnetic fields. The electron collision frequency, on the other hand, does not change as rapidly with increasing temperature as the growth of free carrier electrons. As a consequence, the overall conductivity of silicon generally rises with temperature as shown in Fig. 2-1. This figure illustrates conductivity as a function of temperature for several doped levels of silicon. An initial loss in mobility, due to an increase in collision frequency, causes conductivity to drop until the elevation of electrons to the conduction band overcomes this loss. Eventually the intrinsic carrier concentration rises to such a level that it dominates the doped levels, and the lines merge together. The only potential problem is that the initial decrease may create a coupling valley that inhibits heating to a rather low temperature, $\sim 150^{\circ}\text{C}$. Once past this valley, however, rapid heating can occur via thermal runaway as the electron density and conductivity increase. The temperature of the wafer continues to increase until the electromagnetic power absorbed is balanced by the radiation losses.

An approximate solution for the power absorption efficiency of a silicon wafer can be obtained analytically through a self-consistent solution of both Ampere's and Faraday's laws. The result is derived in Appendix A and follows the methodology of [17]:

$$P_{abs} = \frac{\pi a^2 t_w^3 / \delta_{EM}^4 \sigma}{1 + (t_w / \delta_{EM})^4} H_o^2 \text{ Watts} \quad (2-6)$$

$$\delta_{EM} = \sqrt{2 / \omega_o \mu \sigma} \text{ meters} \quad (2-7)$$

includes the effectiveness of electromagnetic field penetration and depends on a parameter referred to as the skin depth parameter δ_{EM} . In eqn. (2-6), "a" represents the wafer diameter (πa^2 = surface area), t_w the wafer thickness, σ the conductivity, μ the material permeability, and H_o is the magnitude of the magnetic field. This equation simultaneously includes the heating rate physics of eqn. (2-1) and the field penetration physics of eqn. (2-4). It also accommodates the finite thickness of the wafer stack.

Equation (2-6) is plotted in Fig. 2-2 at 13.56 MHz, 2.45 GHz, and 110 GHz. For low conductivity materials (i.e. quartz) the skin depth approaches infinity, indicating complete penetration of the fields through the wafer with very little absorption. As conductivity increases, skin depth decreases and more of the field energy is ohmically absorbed in the material. Eventually the conductivity increases to such a level that the skin depth becomes very small, indicating that most of the electromagnetic field is reflected (e.g. copper) and heating can only occur in the very thin skin depth layer.

A final analysis, plotted in Fig. 2-3, incorporates the effect of temperature on conductivity and shows the absorption efficiency for silicon solely as a function of temperature. The predicted increase in absorbed power at higher temperatures (900°C –

1000°C) is apparent for the radio and microwave frequencies. However, in the millimeter range, the skin depth becomes so small that the heating efficiency actually decreases. While still slightly higher than the microwave radiation, the loss in absorption efficiency indicates that millimeter wave heating is not necessarily more effective for silicon processing than lower frequency solutions.

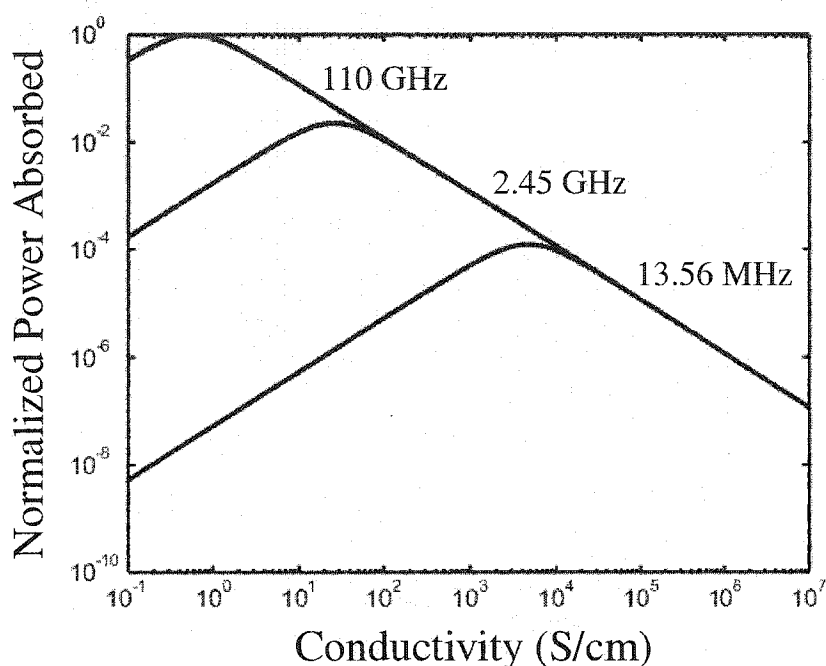


Fig. 2-2. A first order solution for the power absorption efficiency of a silicon wafer obtained through a self-consistent solution of both Ampere's and Faraday's laws.

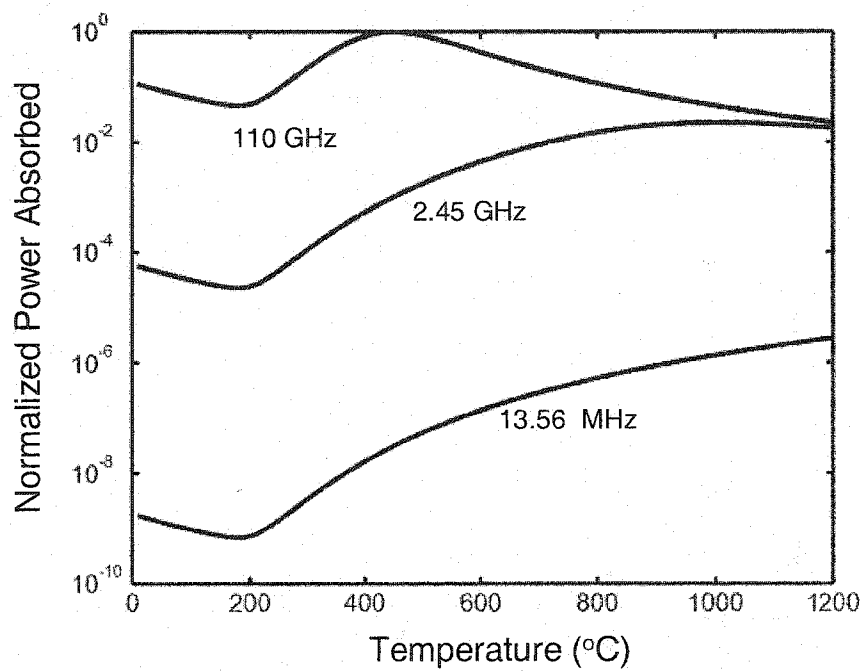


Fig. 2-3. Overall absorption efficiency of electromagnetic radiation into silicon for the radio, microwave, and millimeter frequency regimes.

CHAPTER 3: EXPERIMENTAL APPARATUS

3.1 Microwave Heating

In the microwave regime, electromagnetic induction heating (EMIH) was performed in a resonant cavity of radius 17 cm (Fig. 3-1). The height of the cavity was left adjustable to tune in specific modes, and a radial tuning stub helped to minimize reflected power. Although several modes were found that resonated at 2.45 GHz, the dominant TM₁₁₁ and TM₀₁₁ modes were primarily used. All experiments were strictly single mode, but with additional power sources multi-mode

cavities could be constructed. Up to 3000 watts at a frequency of 2.45 GHz was available from a magnetron source, but to date no more than 1500 watts has been necessary. Recent modifications allow for specific control over the chamber ambient, especially the concentration of oxygen during the anneal process. Fig. 3-2 illustrates the empty cavity electric flux patterns (both TM₁₁₁ and TM₀₁₁ modes), and Fig. 3-3 shows the relative intensity of the axial directed electric flux in an empty cavity operating in resonance.

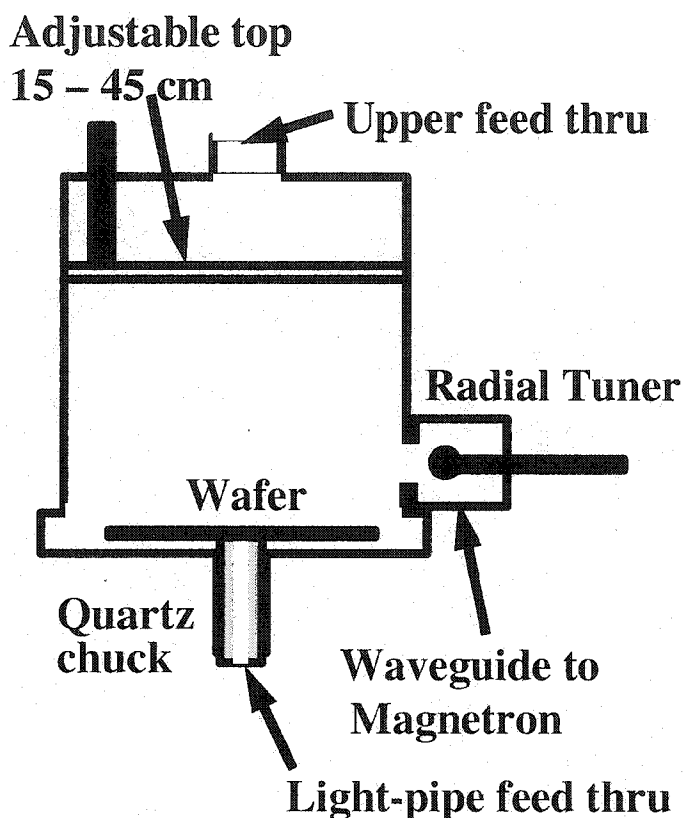


Fig. 3-1. Experimental Apparatus: Resonant cavity for microwave heating.

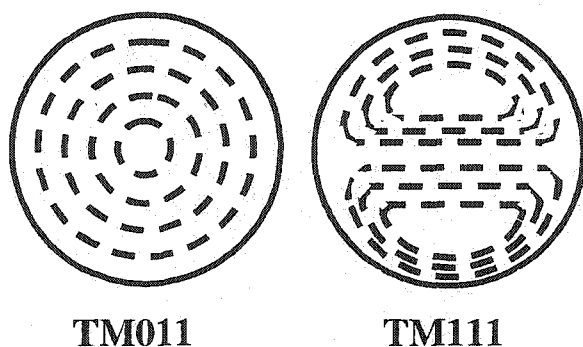
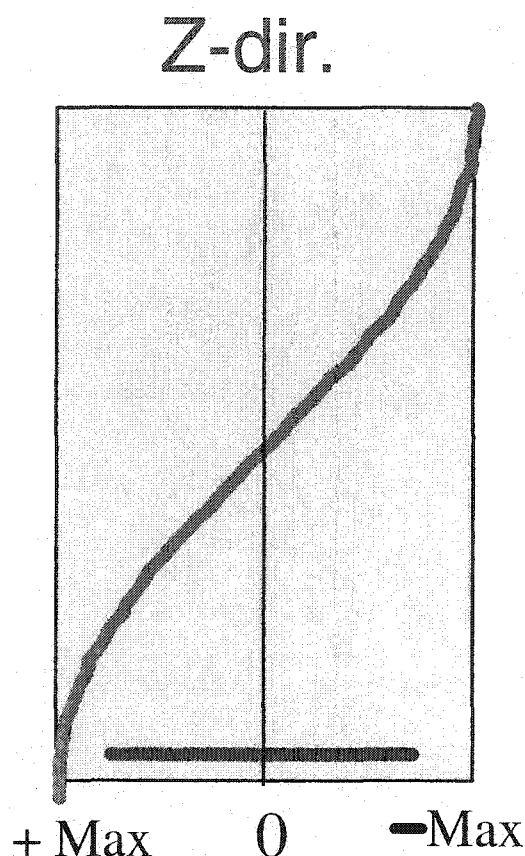


Fig. 3-2. Electric flux patterns, transverse to the wafer surface, for an empty cavity

Fig. 3-3. Relative intensity of the axial directed electric flux in an empty resonant cavity.



The wafer was supported using several different configurations. Figure 3-4 illustrates these wafer support structures. The steady-state uniformity studies were performed using a quartz cylinder to support the wafer, Fig. 3-4a. The diameter of the quartz cylinder that contacted the center of the wafer was 2 cm. The steady-state uniformity studies of the wafer during rotation were performed using the quartz structure shown in Fig. 3-4b. In this configuration, only the outer 1 mm of the silicon wafer actually contacted the wafer holder. Finally, when the small samples were heated for the ultra-shallow junction studies, a 75 mm wafer was placed on several glass slides. The glass slides contacted the wafer only at the edges and were spaced equidistant from each other, as illustrated in Fig. 3-4c. The silicon sample that was annealed was placed in the center of this silicon wafer.

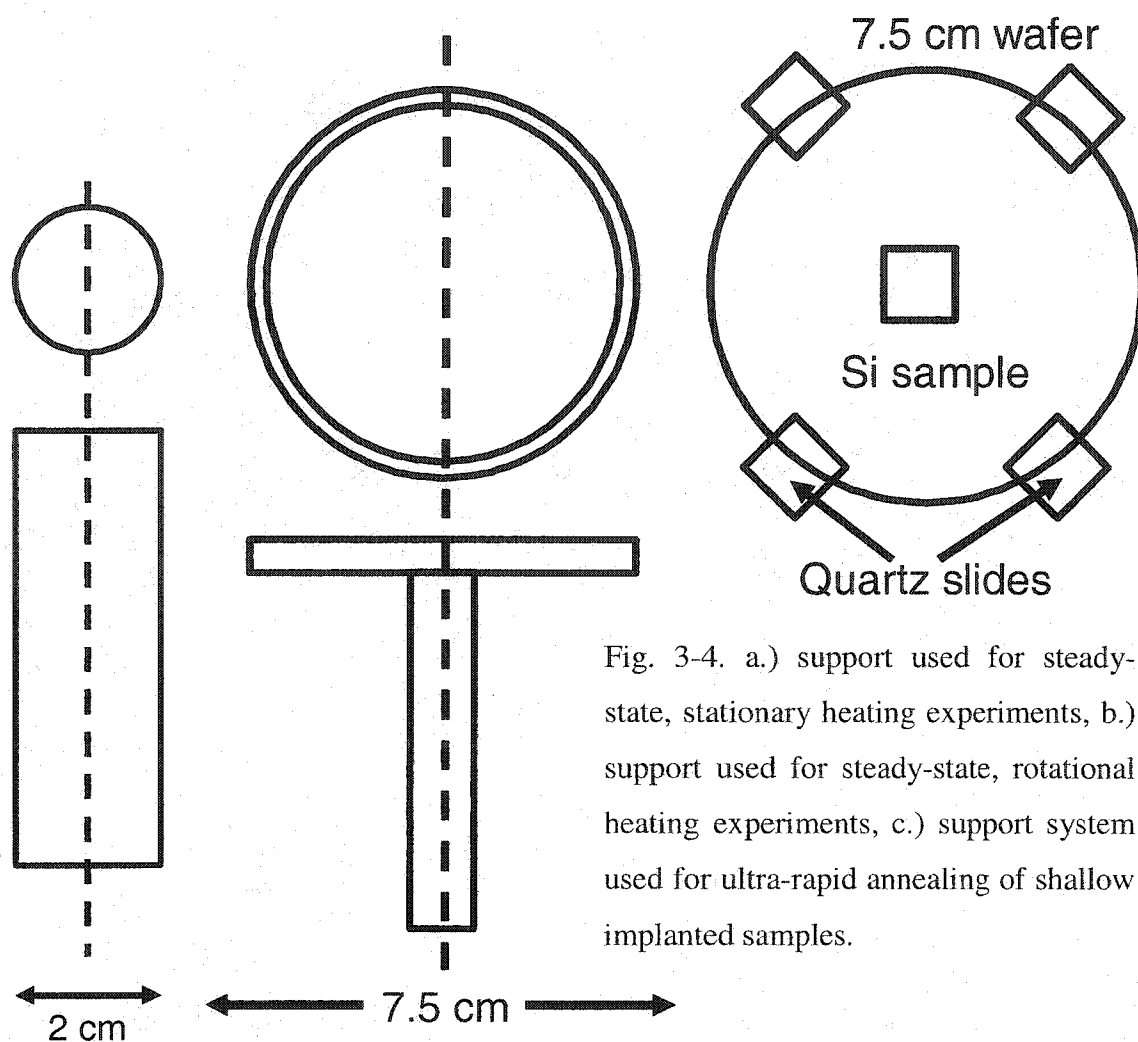


Fig. 3-4. a.) support used for steady-state, stationary heating experiments, b.) support used for steady-state, rotational heating experiments, c.) support system used for ultra-rapid annealing of shallow implanted samples.

3.2 RF Heating

Since EMIH in the RF regime was not practical in a resonant cavity, an alternate approach, illustrated in Fig. 3-4, employed a spiral copper antenna to excite RF magnetic flux. Up to 1000 Watts from a fixed frequency, 13.56 MHz power supply was provided through an L-type matching network. The wafer was positioned on ceramic support rods 2.5 cm below the coil, placing the wafer in the extreme near field of the antenna. A resistive heating element ran underneath the ceramic rods, and the wafer could be pre-heated to 150°C.

3.3 Temperature Measurement

Temperature measurement in the presence of intense electromagnetic fields requires careful techniques [18,19]. An optical pyrometer or light-pipe – calibrated to the specific emissivity of silicon in the temperature range of interest (800-1100°C) – has been shown to accurately measure the temperature without perturbation of the field patterns [18,19]. The light-pipe can be fed through the bottom of the chamber to obtain localized spatial measurements while the pyrometer can remotely view the wafer from outside the chamber to obtain an areal temperature average.

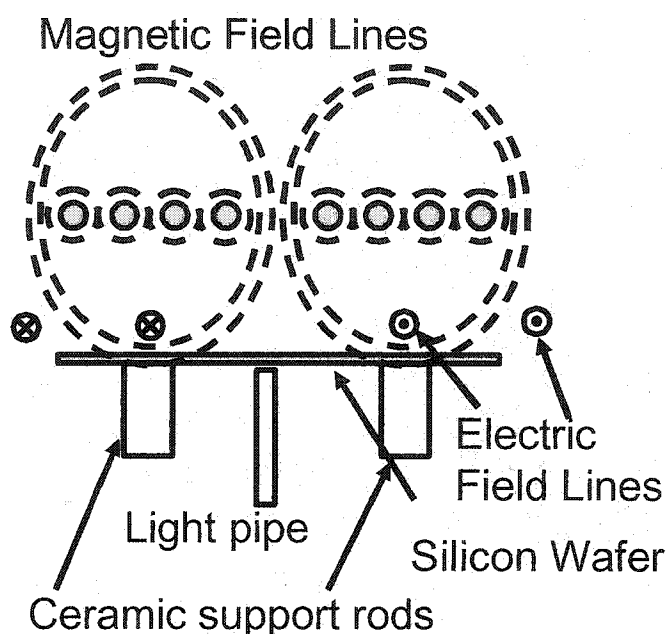


Fig. 3-5. Experimental Apparatus: Spiral coil design used for RF heating.

CHAPTER 4: HEATING RESULTS

Several wafers, whose conductivity ranged from intrinsic ($\rho = 500 \text{ } \Omega\text{-cm}$) to highly doped ($\rho = .001 \text{ } \Omega\text{-cm}$), and whose diameters ranged from 2.5 cm to 10 cm, were heated in the microwave and RF system. Rapid heating in the microwave system was successful without the need to pre-heat the wafer. In the RF system, wafers with a bulk resistivity of 20 ohm-cm or greater required pre-heating to 150°C. This was not unexpected since the higher frequency microwave radiation was expected to heat silicon more efficiently, and move through the coupling valley discussed in Chapter 2, than the lower frequency radio wave radiation. The heating rate in both systems, once the temperature exceeded 500°C, was independent of the room temperature wafer conductivity. This was not unexpected. Once the wafer temperature exceeds 500°C the number of free carriers (electrons and holes) in silicon is dominated by the intrinsic carrier concentration [20,21] and is therefore independent of doping level. This free carrier concentration determines wafer conductivity, which is the key parameter in electromagnetic power absorption efficiency.

4.1 RF Heating

Steady state temperature results, as a function of power and wafer radius, Fig. 4-1, illustrate a significant drawback to the RF heating configuration selected for this study. A significant ($\sim 250^\circ\text{C}$) temperature gradient exists in the radial direction (center cool, edge hot). An examination of the net flux patterns, generated by the antenna/wafer location in Fig. 3-4 reveals the reason for this temperature gradient. Because the silicon wafer is a semiconductor, a fraction of the radiation is absorbed in the silicon, a fraction of the radiation

is reflected by the silicon, and a fraction of the radiation is transmitted through the silicon. At temperatures above 500°C, very little of the radiation is transmitted through the silicon; however, a significant fraction is reflected. At the center of the wafer, this reflected radiation can be modeled as radiation reflected from a semi-infinite plane. The incident electromagnetic flux induces

current to flow in the semi-infinite silicon plane. These currents re-induce an electromagnetic flux that opposes, and thereby reduces, the intensity of the incident flux. The result is that the center of the wafer experiences a significant reduction in flux density due to the presence of the induced current. Near the edges of the wafer, however, the semi-infinite plane model is no longer valid. The presence of a boundary, beyond which no reflected radiation exists (neglecting fringing effects), significantly diminishes the field reduction effect. As a result, the electromagnetic flux density is higher near the edges than at the center. Because the induced currents ultimately depend upon the net flux density (field imposed by the antenna currents minus the field arising from the induced wafer currents), a non-uniform, azimuthally symmetric, radial heating pattern occurs.

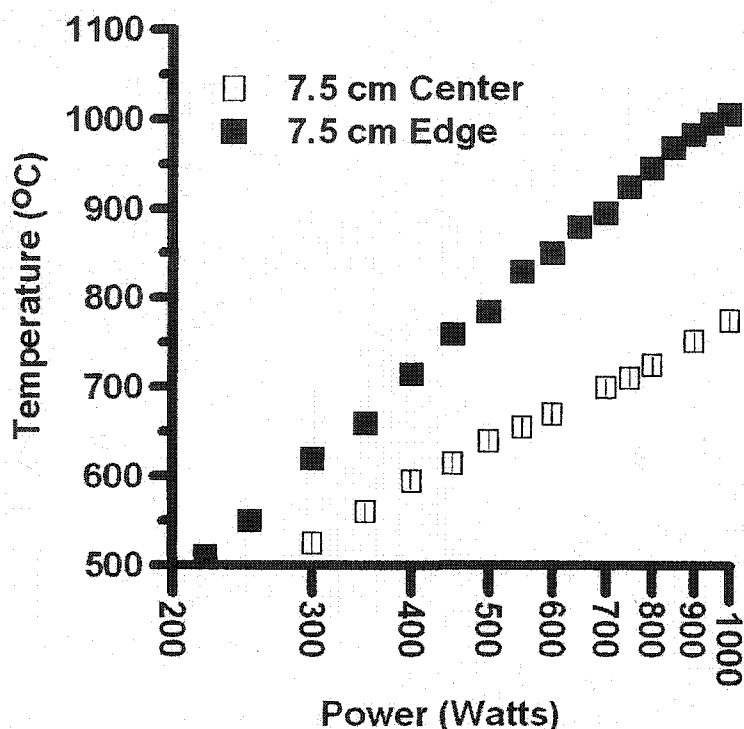


Fig. 4-1. Experimental temperature vs. power results at the center and edge of a 7.5 cm silicon wafer heated with the RF spiral antenna illustrated in Fig. 3-5.

A semi-analytical solution for this heating profile was obtained by assuming quasi-static fields (which is permissible when the radiation wavelength is long compared to the wafer thickness) using mixed boundary value conditions [22] to define the electric flux density inside and outside the radius of the wafer. The solution of the current density is: with the net absorbed power density as:

$$J(r) = \frac{J_o}{\sqrt{1-(r/a)^2}} \quad \text{A/cm}^2 \quad (4-1)$$

$$S(r) = \left(\frac{V_o}{\pi} \right)^2 \frac{\sigma}{a^2 - r^2} \quad \text{W/cm}^2 \quad (4-2)$$

J and S are the current and power density as a function of radius, r, respectively. J_o is the current density at the center of the wafer. The wafer radius is "a", and σ is the electrical conductivity of the silicon. V_o is the quasi-static electric potential that exists on the electrically insulated wafer (i.e. line integral of the electric field). Because the conductivity is high, and the radiation frequency is low, the potential across the wafer is relatively constant.

A steady-state, semi-analytical temperature distribution can be extracted from the power distribution predicted by eqn. 4-2, and the data in Fig. 4-1. Steady-state conditions dictate that the power absorbed by the wafer must equal the power lost. The data in Fig. 4-1 indicates that steady-state temperature depends on the $1/4$ root of power ($T_{\text{wafer}} \sim P_{\text{EM}}^{1/4}$ or $P_{\text{lost}} \sim T_{\text{wafer}}^4$), indicating a power loss mechanism dominated by radiation losses. This is typical for a material whose temperature exceeds 500°C. A first-order approximation for the steady-

state, radial temperature distribution is obtained by applying this $\frac{1}{4}$ order proportionality to eqn. 4-2:

$$T(r) = A \left(\frac{\sigma}{a^2 - r^2} \right)^{\frac{1}{4}} \quad ^\circ\text{C} \quad (4-3)$$

and replacing the various constants with a single fitting parameter "A", which is empirically extracted from Fig. 4-1. The result, Fig. 4-2, represents an accurate depiction of the actual heating pattern.

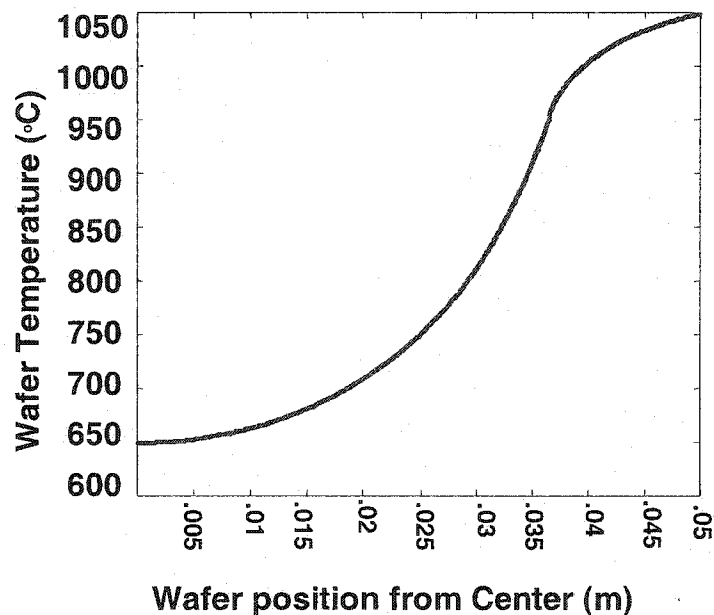


Fig. 4-2. Fitted model for the radial temperature non-uniformity which occurs when heating is performed with the RF spiral antenna illustrated in Fig. 3-5. The wafer diameter is 100 mm, and the input power is 1000 watts.

4.2 Microwave Heating

The temperature transient for an intrinsic, 7.5 cm diameter wafer, heated in the microwave system with 1000 watts of power, is shown in Fig. 4-3. The measured heating rate of 125°C/sec is comparable to the heating rates achieved in commercial lamp-based RTP systems [23]. Steady state temperature results, as a function of power and wafer radius, are shown in Figs. 4-4 and 4-5. The heating efficiencies of the two microwave modes, TM111 and TM011, and the loading effect for heating different diameter wafers are distinguished in Fig. 4-4. The fourth root dependence between temperature and net power absorbed, which must equal the net power lost, again indicates that radiation is the dominant loss mechanism. Radiation losses directly depend on the surface area of the material. The surface area of a silicon wafer is dominated by the "top" and "bottom" surfaces, with little contribution from the thin edges. Stacking numerous wafers on top of each other increases the total volume of material without significantly adding to the overall surface area of the material. As a result, microwave heating has the potential for batch processing. This is illustrated in Fig. 4-5 where, for a given input power, the same temperature is achieved for simultaneously heating several wafers or a single wafer.

Pictures of the wafer taken during microwave heating, Fig. 4-6 and 4-7, indicate that the temperature pattern is not uniform across the wafer. Heating in the TM011 mode results in an edge-hot, center-cool radial temperature pattern. Heating in the TM111 mode results in an hourglass temperature pattern. A quantitative assessment of this temperature distribution, Fig. 4-8, determined through solid state reactions as outlined in [18], confirms this non-uniform temperature distribution. Heating in the TM111 mode results in a temperature non-uniformity of approximately 13.6%, and rotating the wafer at 60 rpm reduces the temperature

non-uniformity to approximately 3.6%. These temperature patterns closely resemble the transverse electric flux patterns [16] that are expected from these given modes, indicating that the electric flux distribution may have a strong impact on the overall heating pattern. This presents a significant drawback to microwave heating in a single mode resonant cavity, and this problem was studied further through simulations.

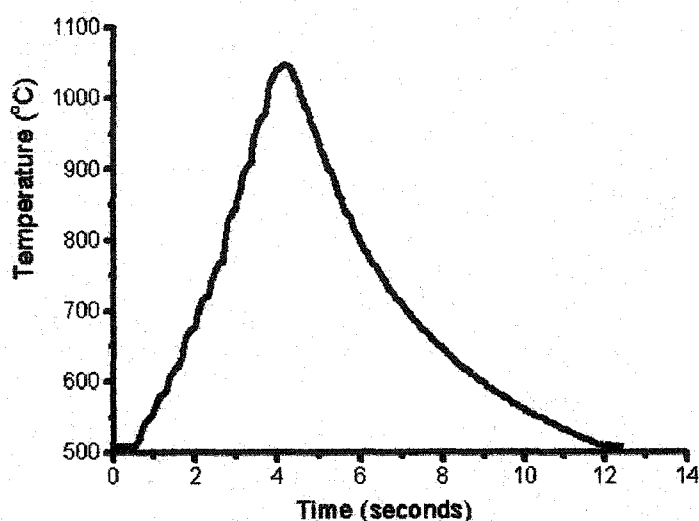


Fig. 4-3. The heating transient of an intrinsic, 7.5 cm diameter wafer heated at 1000 Watts in the microwave system. The measured heating rate is 125°C/sec.

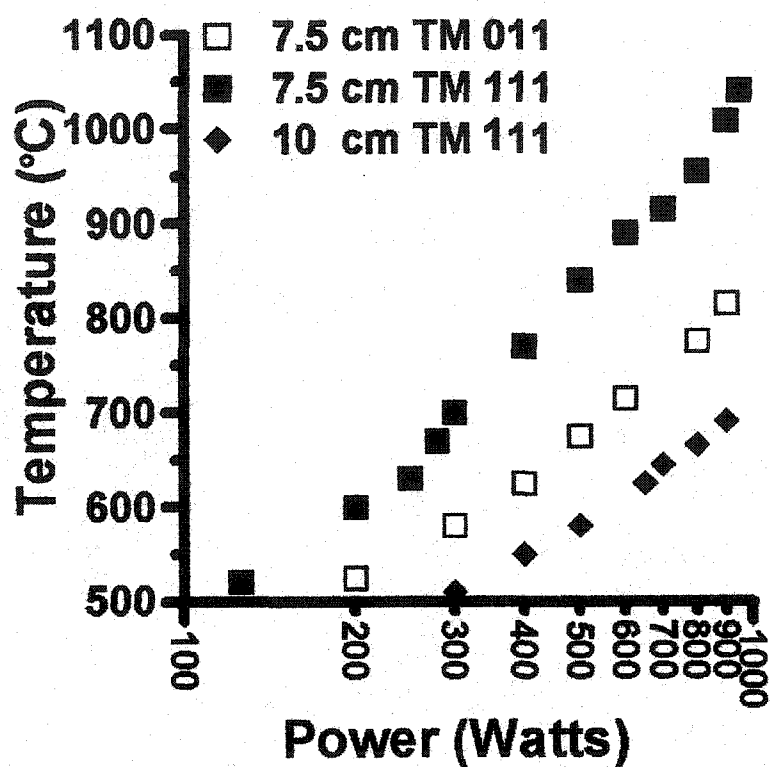


Fig. 4-4. Semilog plot of temperature vs. power for the microwave system.

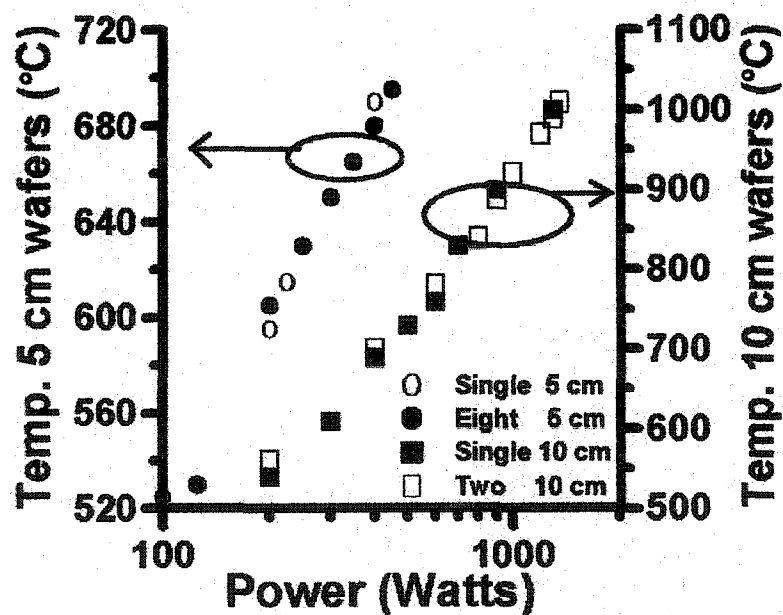
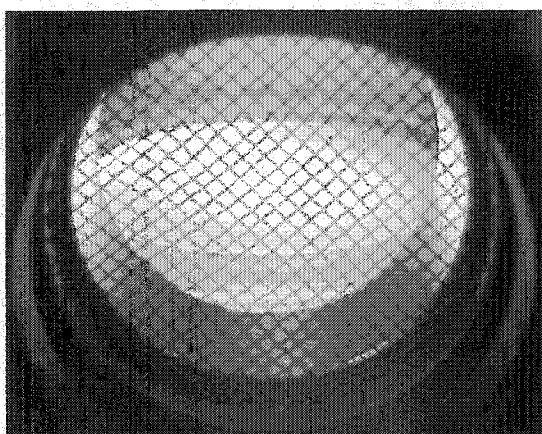
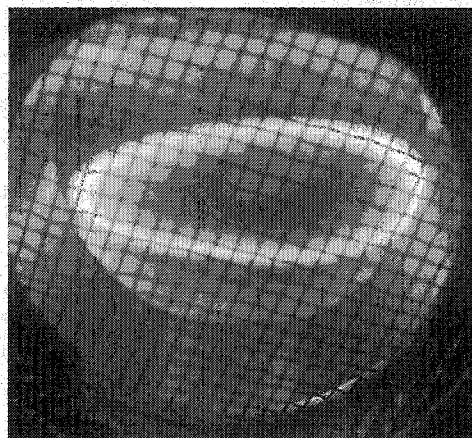


Fig. 4-5. An illustration of the potential that the microwave cavity has for batch heating.



TM 111 rotating



TM 011 rotating

Fig. 4-6. Pictures of the wafer being heated in the TM111 and TM011 mode while the wafer is rotating at 120 revolutions per minute.

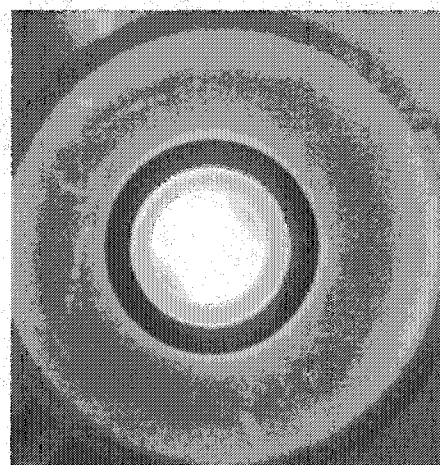
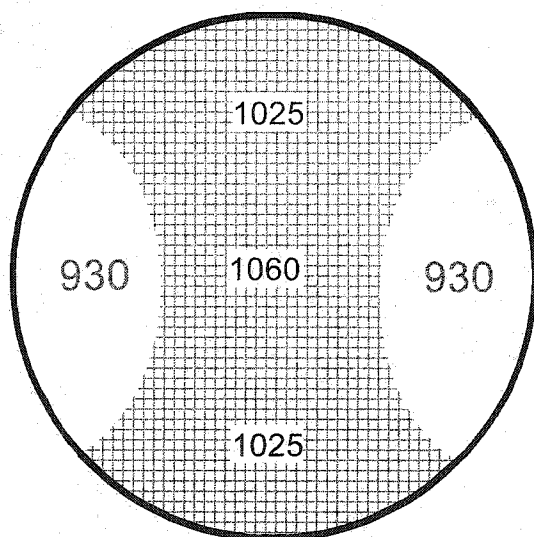
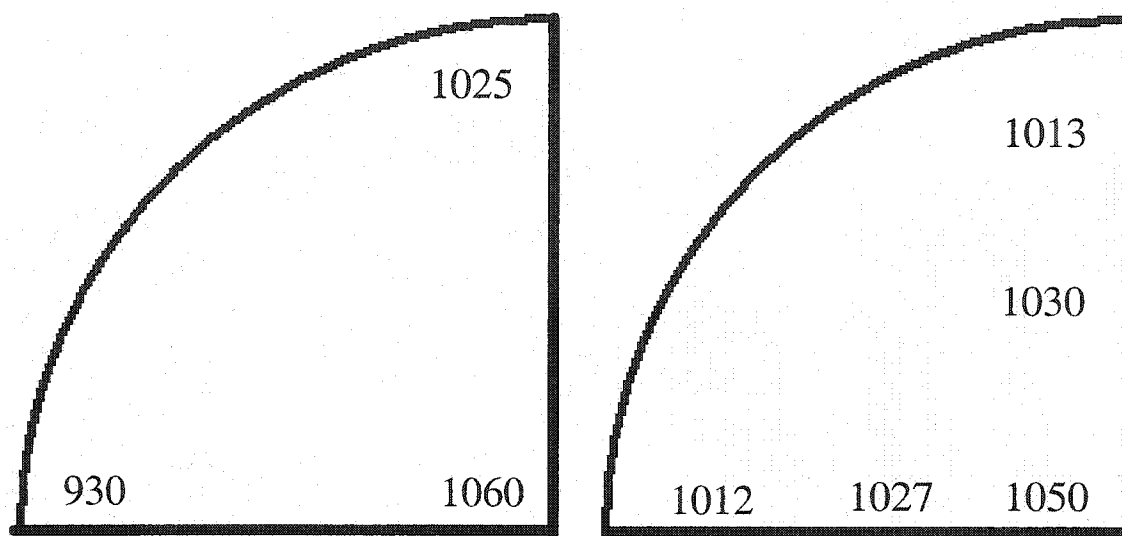


Fig. 4-7. Top down picture of the wafer being heated in the TM111 mode along with quantitative temperature measurements as determined by solid-state reactions.



Stationary wafer, NU~12%

Rotating wafer, NU~3.6%

Fig. 4-8. Uniformity was determined through solid-state diffusion reactions. Wafers were heated with 1000 Watts. Some power loss due to increased reflection during rotation.

4.3 Microwave Simulations

Finite difference time domain (FDTD) simulations were run to examine the electric and magnetic flux intensity spatially within the wafer and to evaluate the subsequent effect on heating uniformity. The rationale for utilizing FDTD to simulate the electromagnetic (EM) fields can be found in [24]. Equations 4-1 and 4-2, derived in Chapter 12 of [24], break the chamber down into a two dimensional (2-D) grid and simultaneously solve for the radial, r , axial, z , and azimuthal, ϕ , electric, E , and magnetic, H , field components in each cell of the 2-D radial coordinate system (radial and axial dimensions). The azimuthal dependence is then obtained by taking advantage of the physical symmetry of the cylinder to assemble the Fourier components around the azimuthal direction as outlined in a typical electromagnetics textbook [16]. In eqns. 4-1 and 4-2, μ is the permeability, ϵ is the permittivity, m is the mode

number, the superscript n refers to the time step, and the subscript 1 or 2 or 0 represents spatial positioning with 0 being the center point and 1 or 2 representing the two surrounding grids necessary to carry out the numerical differentiation.

$$\frac{\partial}{\partial t} E_r^{n+1}(r_o, z_o) = \frac{1}{\varepsilon \Delta z} [H_\phi^n(r_o, z_1) - H_\phi^n(r_o, z_2)] - \frac{m}{\varepsilon r_o} H_z^n(r_o, z_o) \quad (4-1a)$$

$$\begin{aligned} \frac{\partial}{\partial t} E_\phi^{n+1}(r_o, z_o) &= \frac{1}{\varepsilon \Delta r} [H_z^n(r_1, z_o) - H_z^n(r_2, z_o)] \\ &+ \frac{1}{\varepsilon \Delta z} [H_r^n(r_o, z_2) - H_r^n(r_o, z_1)] \end{aligned} \quad (4-1b)$$

$$\begin{aligned} \frac{\partial}{\partial t} E_z^{n+1}(r_o, z_o) &= \frac{r_2}{\varepsilon (r_2^2 - r_1^2)} H_\phi^n(r_2, z_o) - \frac{r_1}{\varepsilon (r_2^2 - r_1^2)} H_\phi^n(r_1, z_o) \\ &+ \frac{m \Delta r}{\varepsilon (r_2^2 - r_1^2)} H_r^n(r_o, z_o) \end{aligned} \quad (4-1c)$$

$$\frac{\partial}{\partial t} H_r^{n+1}(r_o, z_o) = \frac{1}{\mu \Delta z} [E_\phi^n(r_o, z_2) - E_\phi^n(r_o, z_1)] - \frac{m}{\mu r_o} E_z^n(r_o, z_o) \quad (4-2a)$$

$$\begin{aligned} \frac{\partial}{\partial t} H_\phi^{n+1}(r_o, z_o) &= \frac{1}{\mu \Delta r} [E_z^n(r_2, z_o) - E_z^n(r_1, z_o)] \\ &+ \frac{1}{\mu \Delta z} [E_r^n(r_o, z_1) - E_r^n(r_o, z_2)] \end{aligned} \quad (4-2b)$$

$$\begin{aligned} \frac{\partial}{\partial t} H_z^{n+1}(r_o, z_o) &= \frac{r_1}{\mu (r_2^2 - r_1^2)} E_\phi^n(r_1, z_o) - \frac{r_2}{\mu (r_2^2 - r_1^2)} E_\phi^n(r_2, z_o) \\ &+ \frac{m \Delta r}{\mu (r_2^2 - r_1^2)} E_r^n(r_o, z_o) \end{aligned} \quad (4-2c)$$

$$\frac{\rho_T c_T}{k_T} \frac{\partial}{\partial t} T = \left[\frac{1}{r} \frac{\partial}{\partial r} r \frac{\partial T}{\partial r} + \frac{\partial^2 T}{\partial z^2} + \frac{1}{r^2} \frac{\partial^2 T}{\partial \phi^2} \right] + \frac{P}{k_T} \quad (4-3)$$

4.3.1 Quality Factor Calculation

The quality factor, or Q , represents the amount of energy stored in the cavity relative to the amount of input power while the system operates under steady-state conditions. A high Q represents a slow decay of the cavity energy. The cavity Q was extracted from the simulations by driving the simulated cavity to steady-state, removing the current source, and monitoring the total system power, as a function of time. The result, Fig. 4-9, indicates an exponential decay time constant of 5×10^{-9} . Multiplying this time constant by the radiation frequency (1.54×10^{10} radians/second) yields a Q of 77. Because the cavity walls are assumed lossless in the simulations, and because the walls of the actual cavity are lossy, this Q is somewhat higher than the actual Q of the cavity.

A Q of ~ 77 indicates that the cavity should heat materials in a very efficient manner, as is seen in the temperature transient in Fig. 4-3. If the heating were done in a non-resonant system, for example by bouncing plane-wave radiation off the wafer, the heating would not have been as efficient. When radiation is incident on a material only a fraction of the power is absorbed. The remainder of the power is either

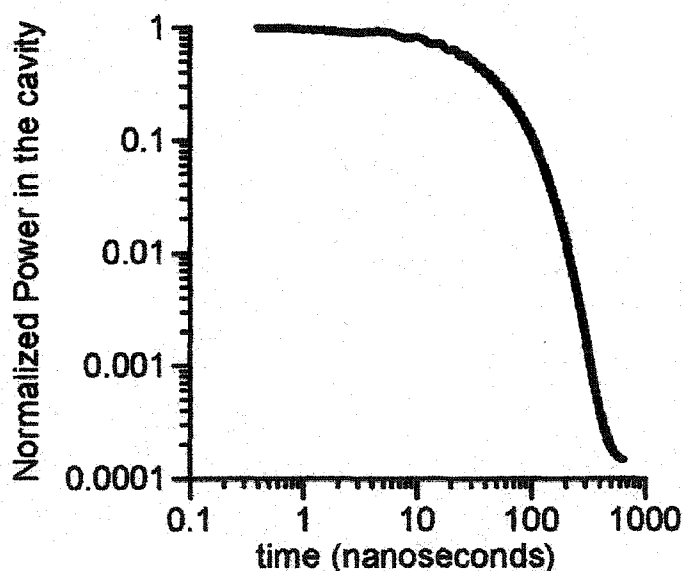


Fig. 4-9. Normalized power as a function of time in the simulated cavity after the current source is removed. The decay rate is exponential with a time constant of 5×10^{-9} /sec.

reflected off of or transmitted through the material. As a result, only a fraction of the input power actually goes into heating the material. In a resonant cavity, however, this non-absorbed power is recycled by the cavity, bouncing from wall to wall until it is again incident on the material. This cycle continues until all of the power is either absorbed by a material, in this case either the silicon wafer or the cavity walls. A system that heats by bouncing plane wave radiation off a material has a Q of 1 since none of the power is recycled. By comparison, the Q of 77 associated with the system used in these experiments indicates that the resonant cavity requires 77 times less power than a system that heats by directing plane wave radiation at the silicon.

4.3.2 Uniformity

In order to extract the steady-state temperature, the simulated electric and magnetic flux intensities were inserted into a thermal FDTD simulation, eqn. 4-3, using a grid matching the EM FDTD simulation. In eqn. 4-3, P is the net power density, k_T is the thermal conductivity, c_T is the specific heat of silicon, ρ_T is the density of silicon, r is the radial direction, ϕ is the azimuthal direction, T is the temperature, and t is time. The net power density is determined by subtracting the absorbed electromagnetic power from the radiation losses. The absorbed power, is calculated for each grid cell of the silicon wafer by multiplying the induced current density, of each grid cell, by the absorbed electric field, for each grid cell. The current density induced in the wafer is determined by multiplying the absorbed electric field, of each grid cell, by the conductivity of the silicon. The absorbed electric field is determined by the iterative solution of Maxwell's equations, which constitutes the core of the EM FDTD simulation. The power losses are determined by calculating the

amount of power radiated from each grid cell of the silicon wafer (i.e. stefan-boltzmann constant multiplied by emissivity multiplied by surface area of the grid cell multiplied by the grid cell temperature taken to the fourth power).

A rigorous assessment of temperature as a function of time can be obtained at each grid cell location on the wafer by stepping the EM power into the thermal code in an iterative, staircase-type fashion whose general methodology is outlined in [25,26,27]. For example, the wafer temperature is set to an initial value of 500°C, and the temperature dependent parameters, specifically conductivity, are calculated for each grid cell. Using these calculated parameters the EM code is run to steady-state (typically several microseconds), and the absorbed power is inserted into the thermal code. The thermal code is then run for the desired time-step (e.g. 1 millisecond), and the temperature dependent parameters of each cell are updated and re-inserted into the EM code. The spatial temperature values are then saved and the process repeated to determine the temperature distribution of subsequent time steps. This iterative process continues until the maximum or target temperature is reached. Despite the approximations involved in this process, the coupling of these two codes works quite well if the time constant to reach steady-state for the EM code is orders of magnitude faster than the time step of the thermal code.

A major drawback to this iterative technique is the enormous computing time involved. A one millisecond time step over a 3 second heating period requires 3000 iterations of the EM code. The EM code took approximately 20 minutes per iteration, requiring 1000 hours of computation time. Because conductivity is the only temperature dependent parameter used in the EM code, and because power absorption as a function of conductivity saturates above 700°C (see Fig. 2-3), these multiple iterations were eliminated in favor of a single

iteration approximation. While the subtleties involved with a spatially varying conductivity are not retained, this methodology still provided an accurate assessment of the heating uniformity under steady-state conditions. Using this simplified methodology, the initial temperature of the wafer was set to 900°C everywhere, with the conductivity calculated accordingly. The EM simulation was then run and the absorbed power transferred into the thermal code. The thermal code was then run until steady-state temperatures values were achieved.

The physical reality of the microwave chamber was adhered to as closely as possible during the simulation. Because the wafer was located 5 mm above the bottom of the chamber, and because each grid was 1 mm in height, five free space grid cells were left between the wafer and the conductive barrier representing the cavity bottom. The quartz chuck upon which the wafer sat was not incorporated into the simulation. The simulated cavity was a perfectly round cylinder of infinite conductivity, but the actual cavity was composed of brass and contained three small view-ports, approximately 2 cm in diameter, which caused negligible perturbations from the idealized cylindrical symmetry. The EM power in the simulation was introduced through a current source located 90 grid cells above the wafer (32 grid cells from the top of the chamber). In the experiments, power was input through an aperture-coupled waveguide at the base of the microwave cavity. Finally, wafer rotation during experimental heating was simulated by averaging the Fourier components used to assemble the azimuthal components of the 2-D code into the full 3-D wafer representation. This averaging provides perfect azimuthal symmetry, as expected from an idealized rotational heating system.

The quality factor, or Q , of the cavity was calculated by running the simulation to steady state, turning off the power supply, and monitoring the decay of the electric flux within the cavity. The Q is an important measure of cavity efficiency since it measures the total energy stored relative to the power input into the cavity. Unlike a "single-pass" system, which heats the silicon by bouncing radiation off the silicon surface,

Contour maps of the axial (normal to the wafer) and transverse (parallel to the wafer) electric field intensity across the wafer surface for the TM111 mode are shown in Figs. 4-10 and 4-11. The resulting temperature map after heating for 0.6 milliseconds, starting from 0°C , and during steady-state heating, starting from 900°C , are shown in Fig. 4-12 and 4-13 respectively. The axial and transverse electric fields for the TM011 mode, along with the simulated temperature, are shown in Fig. 4-14. Because the TM011 and TM111 rotational modes are azimuthally symmetric, the electric field and temperature data for these heating scenarios are represented by a single cut in the radial direction. The axial

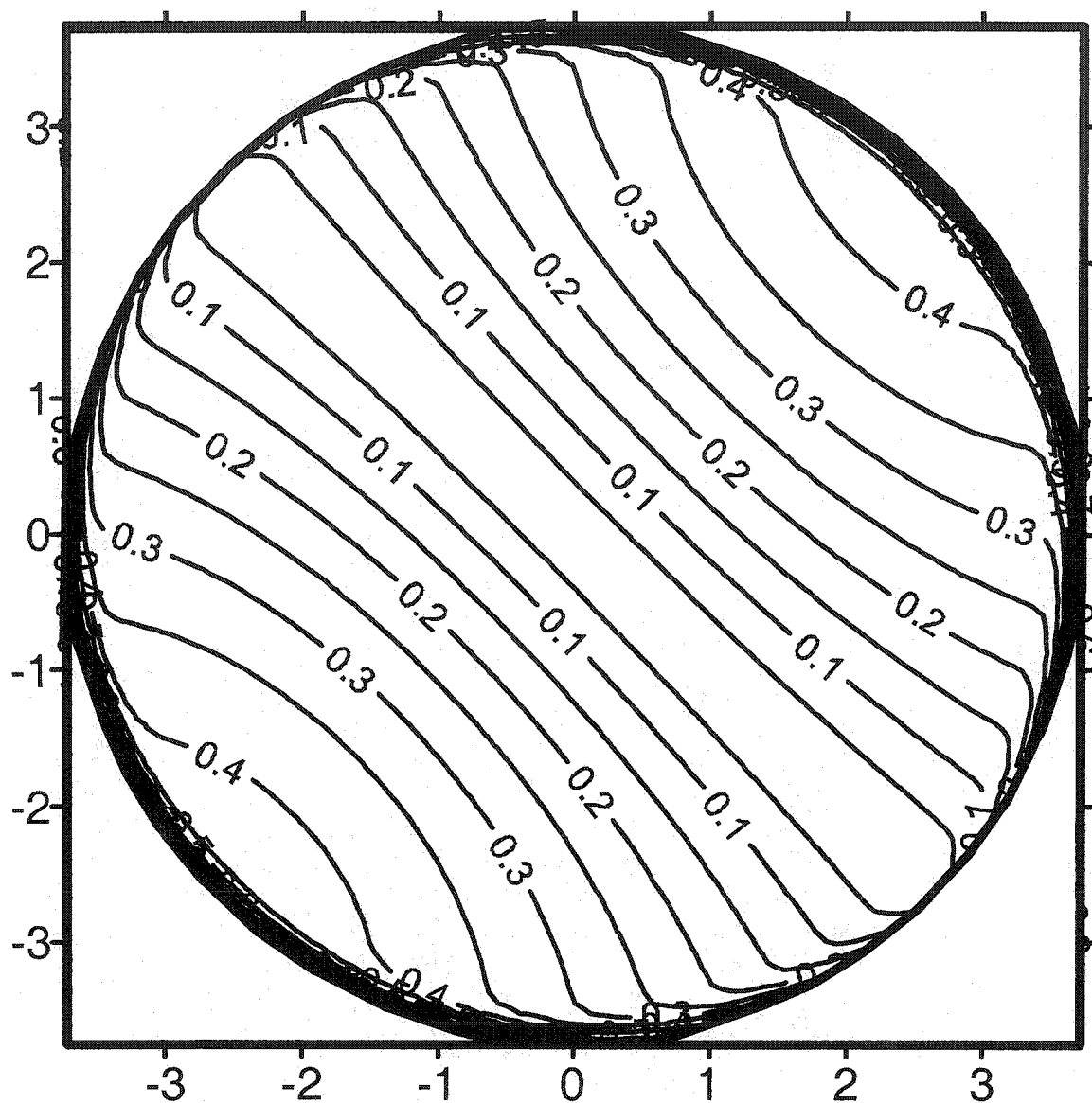


Fig 4-10. Simulated axial electric field (E_z) intensity for a 7.5 cm wafer heated in the TM₁₁₁ mode. The temperature of the wafer is 1050°C, and the field strength has been normalized. The maximum electric field, in the radial range of 3.5 cm from the wafer center, is 4.6 V/cm. The field rises rapidly in the outer 2.5 mm of the wafer, and these values are not accounted for in determining the maximum electric field.

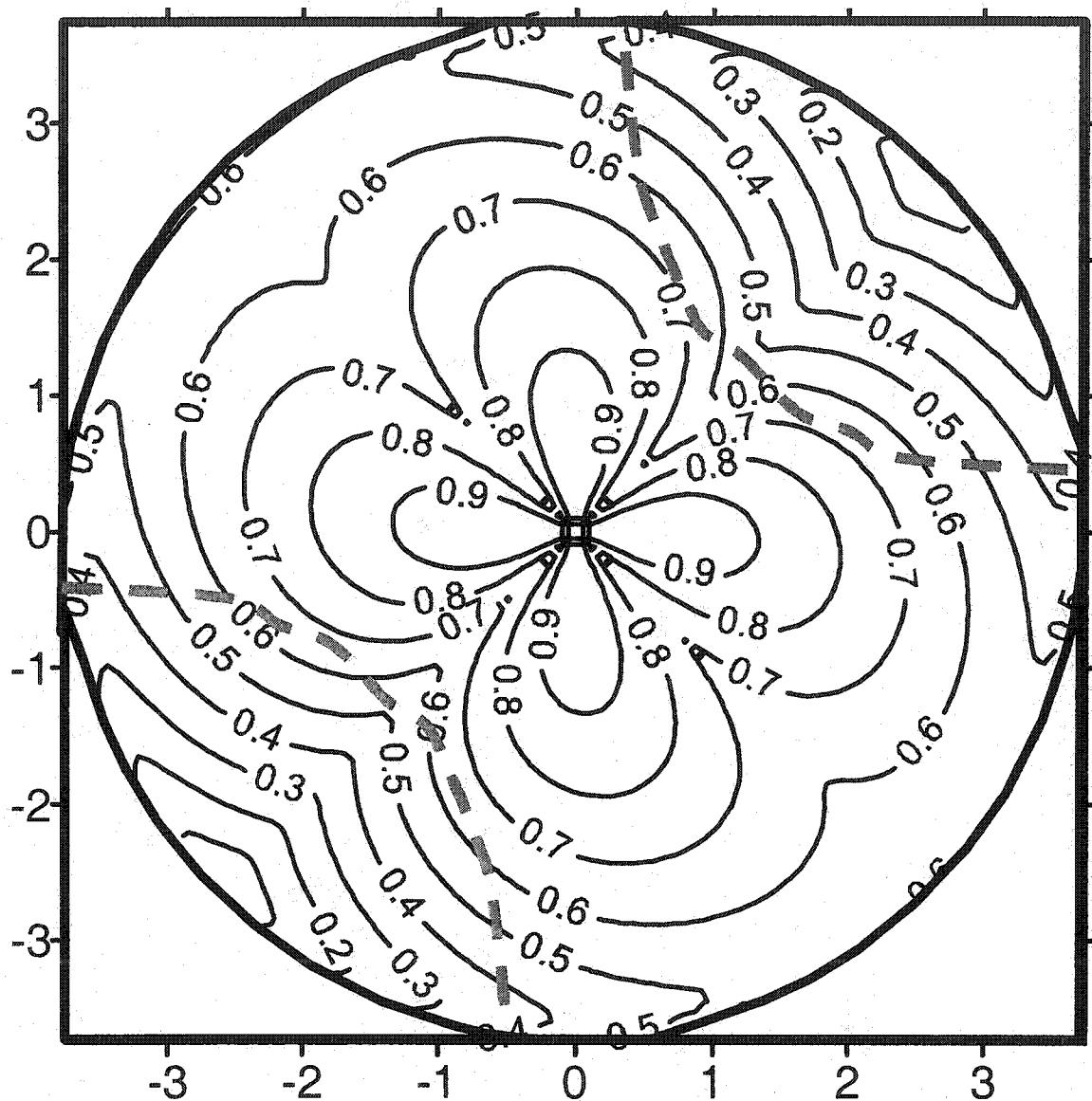


Fig 4-11. Simulated transverse electric field (parallel to the wafer) intensity for a 7.5 cm wafer heated in the TM₁₁₁ mode. The temperature of the wafer is 1050°C, and the field strength has been normalized. The maximum field strength is 146.8 V/cm, approximately 30 times that of the normal component of the electric field.

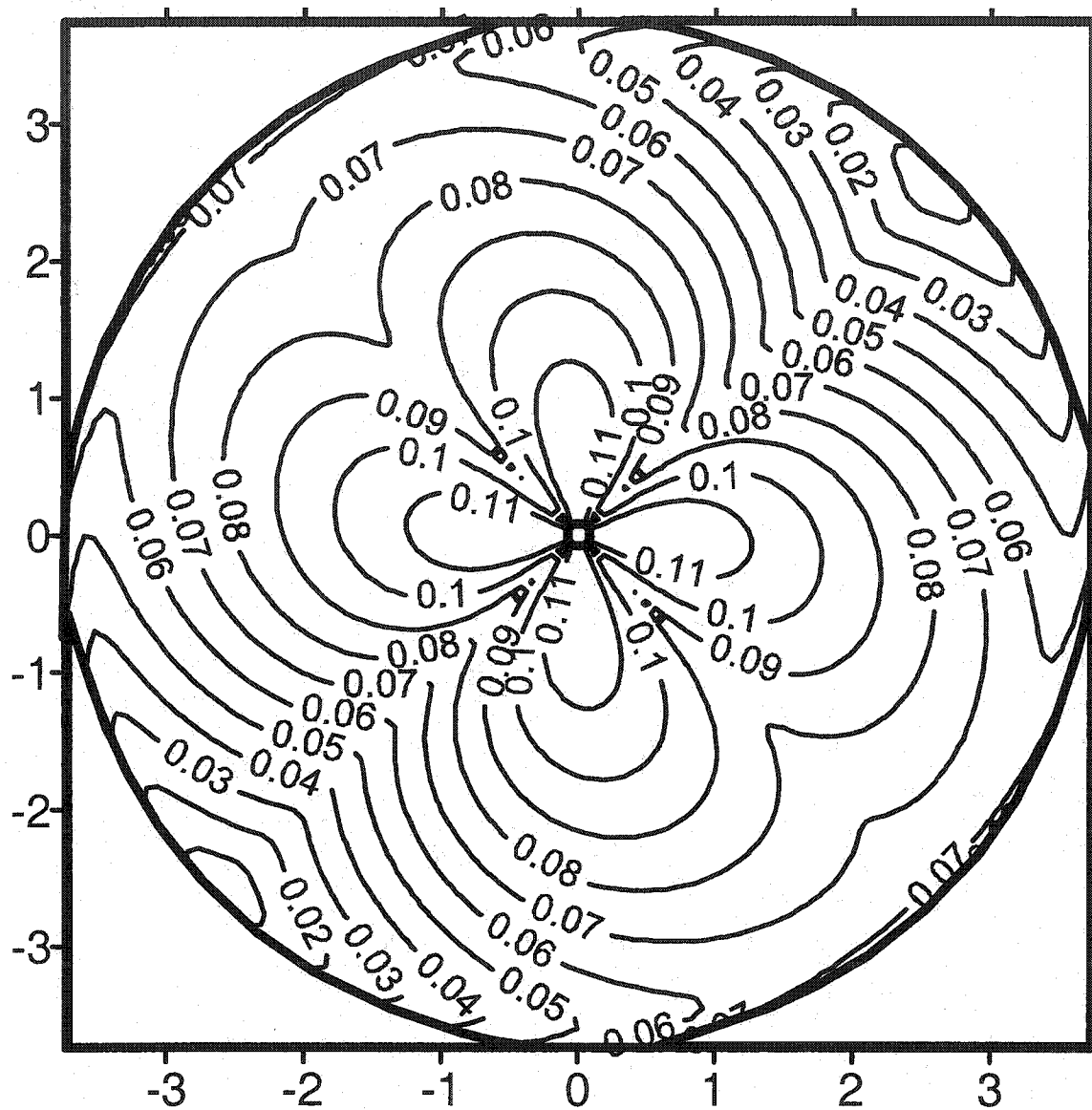


Fig 4-12. Simulated temperature map for a 7.5 cm wafer heated in the TM111 mode for 0.6 milliseconds, starting at 0°C. Max Temp = 0.12°C, Min Temp = 0.02°C

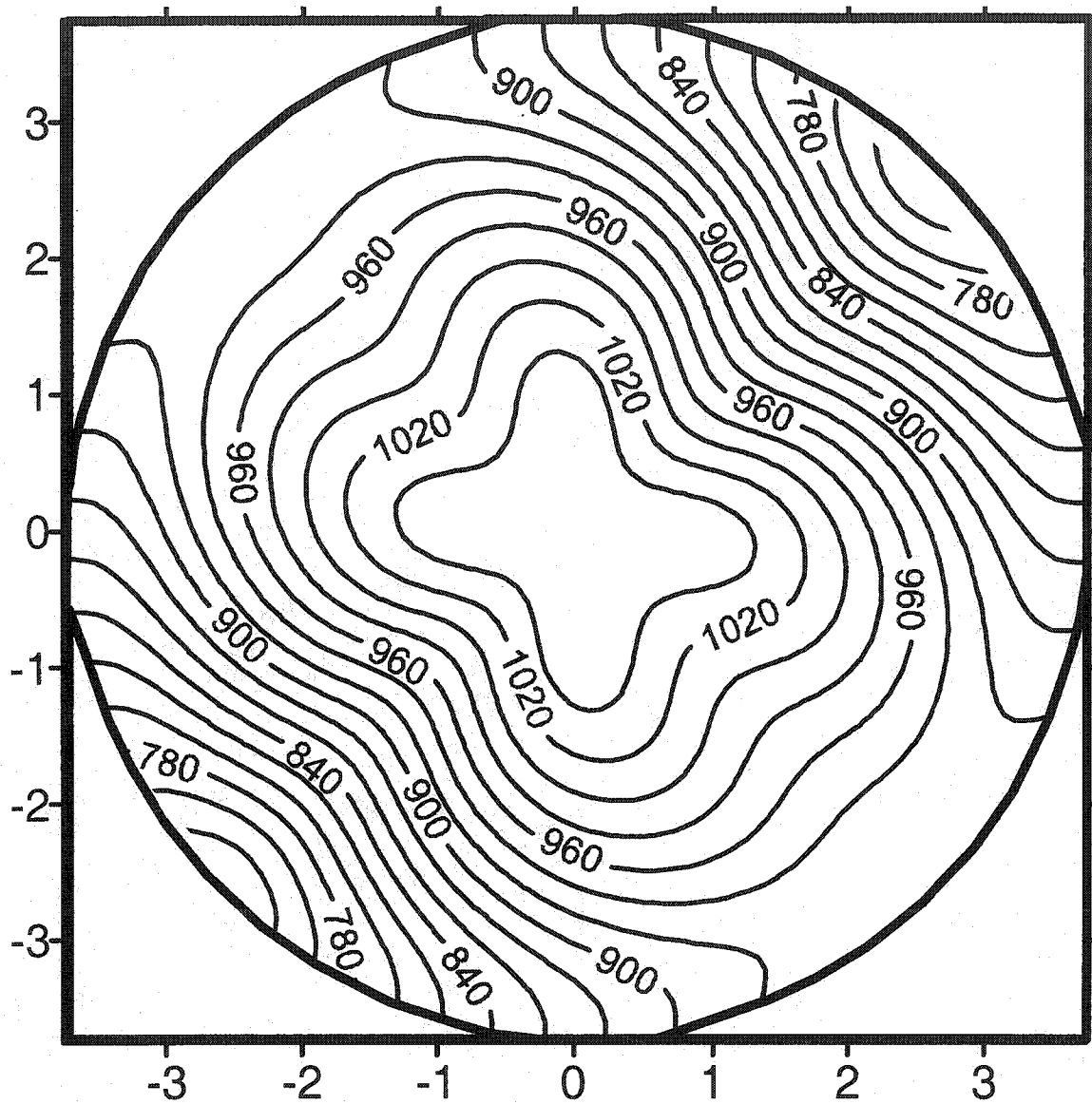


Fig 4-13. Simulated steady-state temperature map for a 7.5 cm wafer heated in the TM111 mode. Max Temp = 1050°C, Min Temp = 740°C

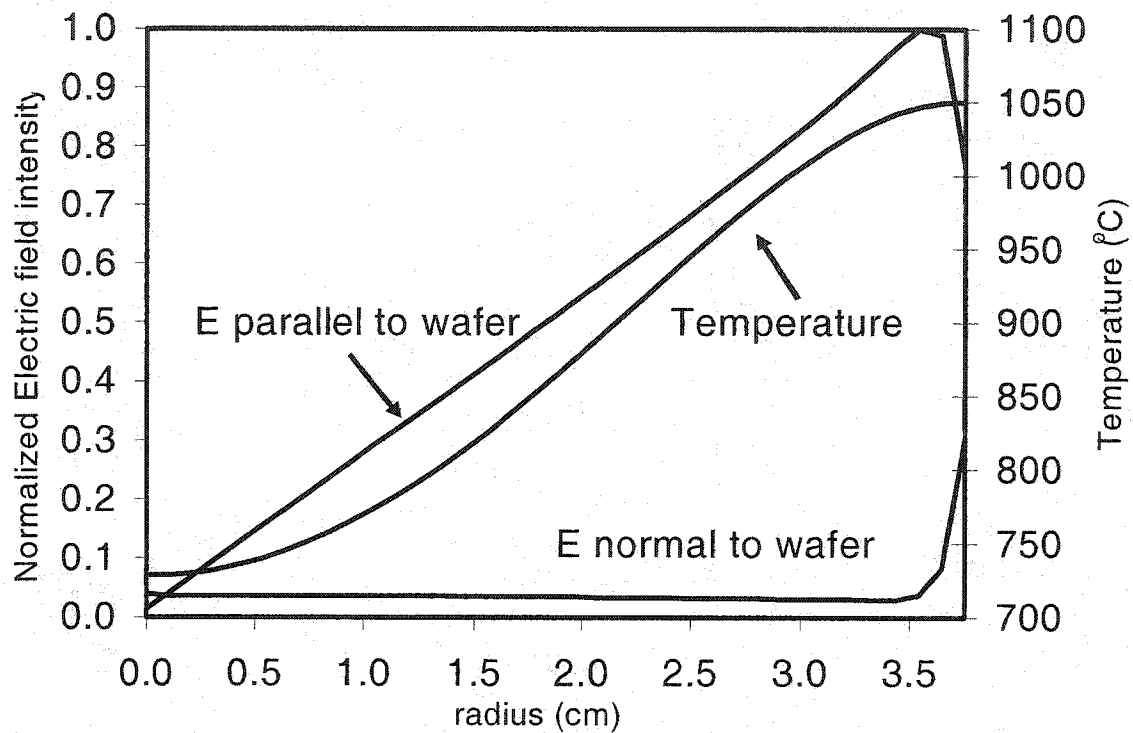


Fig 4-14. Simulated electric field intensities for a 7.5 cm diameter wafer heated in the TM011 mode. Because the field, and therefore temperature, pattern is azimuthally symmetric, a single cut through half the wafer is shown. The maximum temperature of the wafer is 1050°C, the maximum transverse electric field is 11.48 V/cm, and the maximum normal electric field is 3.8 V/cm.

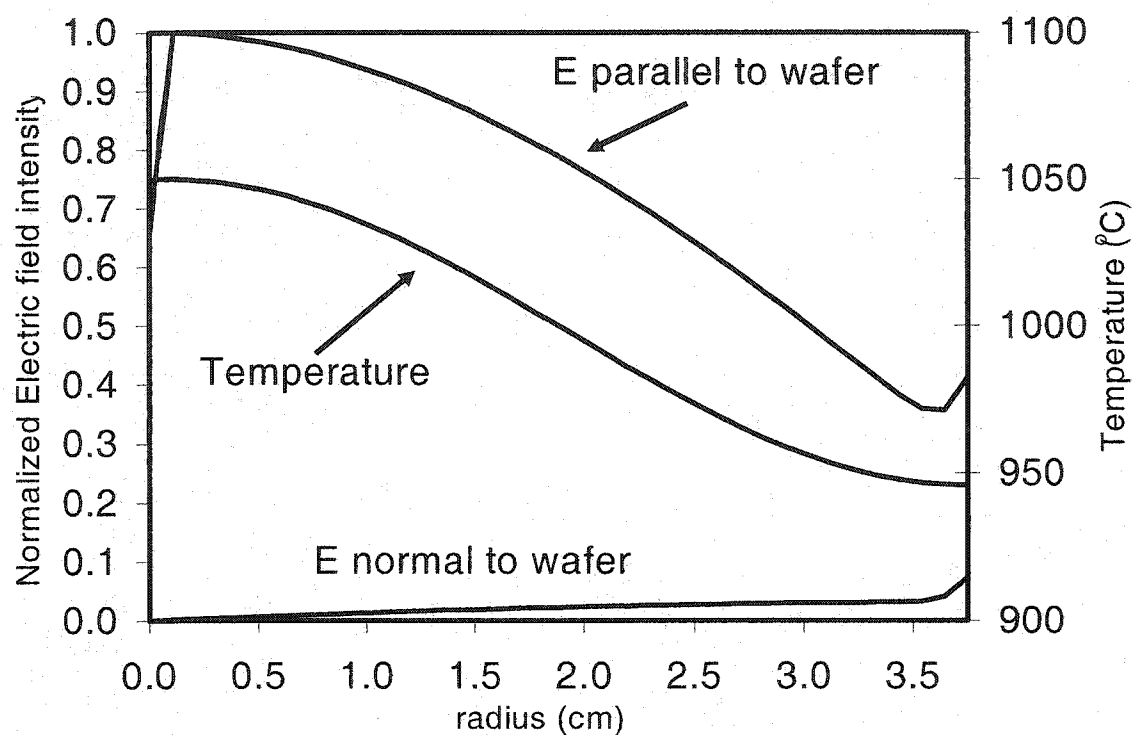


Fig 4-15. Simulated electric field intensities for a 7.5 cm diameter wafer heated in the rotational TM₁₁₁ mode. Because the field (and therefore temperature) pattern is azimuthally symmetric, a single cut through half the wafer is shown. Experimentally measured temperatures are shown as text. The maximum temperature of the wafer is 1050°C, and the maximum electric field is 103.8 V/cm.

and transverse electric fields for the rotational TM111 mode, along with the simulated temperature, are shown in Fig. 4-15. In addition, empirically determined temperatures, at three point points along the wafer radius, are included in Fig. 4-15.

The distinct features of each mode pattern are clear. The static TM111 mode exhibits an hourglass pattern (see dashed lines in Fig. 4-11) with the transverse field intensity strongest within a clover leaf region at the center of the hourglass and the axial field intensity strongest outside the hourglass. The azimuthally symmetric TM011 mode, Fig. 4-14, shows a steep radial gradient with most of the transverse field intensity located at the edge of the wafer while the maximum axial field intensity is focused in the center of the wafer. Note that continuity of the total electric field, within the physical confines of the chamber, means that the transverse and axial electric fields have alternating maximums and minimums. Rotating the wafer averages out, in time, the azimuthal asymmetry of the TM111 mode, eliminating the hour glass pattern effect of the TM111 mode, but a distinct radial gradient (center hot, edge cool) still exists.

The "hot" zones for each mode correspond with the areas of maximum transverse electric field intensity (the approximate complement of the axial electric field), and it is clear that the electric flux intensity transverse to the wafer surface has a dominating impact on the overall temperature pattern. The range in simulated temperature across the wafer (calculated as the maximum temperature minus the minimum temperature divided by the average temperature) is greatest for the TM011 mode, 37%, and non-rotating TM111 mode, 32%. The simulated temperature non-uniformity of the TM111 wafer during rotation improves to 10% as compared to the non-rotating TM111 mode. While this improvement is considerable, it is not enough to generate a commercially viable product.

The empirically determined non-uniformity results are better than those predicted by the simulations for heating in the TM111 mode. The key difference between simulation and experiment is that the minimum temperature at the edge of the wafer is lower in the simulations than in the experiments. In the "cold" spots at the edge of the wafer, the temperature is $\sim 750^{\circ}\text{C}$ in simulations and $\sim 930^{\circ}\text{C}$ in the experiments. Rotating the wafer by 90° , the simulations predict a temperature of $\sim 920^{\circ}\text{C}$ in the hot spot at the edge of the wafer as compared to 1025°C as determined experimentally. While the quantitative temperature assessment, obtained through the solid-state reactions, was performed approximately 1 cm from the edge of the wafer, the simulated temperature calculations accounted for data points all the way to the wafer edge. This discrepancy between experiment and simulation does not hold true for the TM011 mode simulations, which predict a minimum temperature of 730°C in the center of the wafer as compared to a measured minimum temperature of about 725°C .

The overall result is clear: the intensity of the transverse electric field directly affects the final wafer temperature, and any non-uniformity in the transverse electric field intensity across the wafer surface creates a similar non-uniformity in the wafer temperature. The dependence of microwave heating on the intensity of the electric flux transverse to the wafer surface makes the heating uniformity highly susceptible to any non-uniformity in the electric/magnetic field pattern. Therefore, one can expect heating non-uniformities from microwave heating in a single mode resonant cavity.

It is apparent that uniformity is the dominant obstacle to microwave annealing in a resonant cavity since commercially available RTP systems are generally expected to provide less than 1% non-uniformity in sheet resistance. This issue, however, can be overcome by constructing a dual mode cavity in which both the TM111 and TM011 modes simultaneously

power the cavity. The uniformity could be controlled by adjusting the power of the TM₀₁₁ mode (edge hot) with respect to the TM₁₁₁ mode (center hot) until the radial gradient was eliminated. The main drawback to such a system is the requirement of a second microwave power source whose frequency is selected so that both modes are in resonance for a given set of cavity dimensions. This would necessitate a feedback system that adjusted the power of both modes as well as the physical height of the cavity. While complicated and more expensive than the current experimental apparatus, this design is well within the realm of present day technology.

The resonant cavity approach is not the only means by which microwave radiation can be used to rapidly heat silicon wafers. An alternative system involves a multi-mode cavity whose physical dimensions are many times larger than the free space wavelength of the incident radiation. Because no single mode is resonant, dozens of random and unique modes simultaneously deposit energy into the silicon wafer. The effective mode pattern is an average of all of these modes, with the resultant electric field intensity having excellent spatial uniformity across the wafer surface. One such system [28] has reported temperature ramp rates greater than 150°C/sec with a cross wafer non-uniformity much less than 1%. The main drawback to this design is the increased power and thus cost of the microwave power supply. Because the wavelength of the 2450 MHz radiation used in our resonant cavity is 12.25 cm, a system size greater than 1 meter is required to achieve a truly multi-mode configuration. The power densities needed to obtain the same heating rates as those in a smaller single mode system require a much greater input microwave power for the larger system. This requirement for a very large, very high power system can be avoided by going to higher frequencies. Using a high power gyrotron at millimeter-wave frequencies (30 GHz

and above), wavelengths of 1 cm or less can be achieved and much smaller multi-mode cavities are possible, as in [28]. However, the jump from magnetron (currently used for high power 2450 MHz microwave generation) to gyrotron technology incurs a significant capital investment. This capital investment may be offset by performing the heating in batch mode or by moving to a more rapid (millisecond) heating time frame. This shorter time frame is not realistic with current lamp-based annealing technologies and the advantages that come from millisecond heating may offset the additional costs.

CHAPTER 5: ULTRA-SHALLOW JUNCTIONS RESULTS

Advances in rapid thermal processing (RTP) such as optimized target anneal temperatures with heating rates of 450°C/sec, ambient control, pre-amorphization, and the introduction of secondary impurities such as fluorine during the implant have allowed the 90 nm technology node to be satisfied [23]. Scaling beyond this node requires the utilization of novel rapid heating techniques. Microwave annealing is one such novel possibility [18,19,29,30,31]. To assess the effectiveness of microwave annealing for ultra shallow junction formation (USJ) the well-known models detailing the diffusion and activation of boron were re-evaluated.

5.1 Principles of Diffusion and Activation

Efficient ultra-shallow junction formation requires high levels of activation while minimizing diffusion. To this end, it is necessary to establish a basic understanding of boron transport along with the defect reactions that affect both the movement of boron through the silicon lattice and the activation of boron into silicon sites. The diffusion kinetics of boron in silicon are affected by several factors including defect structures, the presence of impurities, and the presence – or lack of – optical illumination. The large number of variables affecting the diffusion of boron have led to the proposal of numerous models to explain the perceived anomalous movement of boron during the early stages of annealing [8 – 13]. While a great deal of thought and research has gone into these models, each is inconsistent with several aspects of the broad range of experimental data available. In fact, some proposed models are based entirely on ab-initio or Monte-Carlo simulations with no comparison to experimental

work [47]. The following discussion presents a self-consistent analysis of boron transport in silicon during the first few seconds of the anneal process.

Numerous discrete atomic encounters affect the diffusion of boron. Diffusive flux, however, is observed in macroscopic terms. Consequently, only the cumulative result of these individual encounters is observed experimentally while the individual mechanisms that effect the movement of boron remain hidden. To counter this obstacle to analysis, several atomic interactions are presented, and the anticipated impact on the net flux of boron is proposed. Through an analysis of the experimental data, including situations where these various factors are either enhanced or suppressed, the role that these individual reactions play on the overall transport of boron through silicon is verified or discounted.

Boron transport through the silicon lattice is dominated by an interstitial mechanism. Interstitial sites are the octahedral shaped – 6 fold coordinated – spaces in the diamond cubic (FCC) silicon structure. Atoms located in interstitial sites, which are not part of a bonded complex with either adjacent atoms or other interstitials, are nominally free to move throughout the silicon. These atoms are known as free interstitials. At high temperatures, the energy available for the random motion of these interstitials is greatly increased, allowing for rapid diffusion through the crystal lattice. Atoms that occupy interstitial sites but are also part of a defect complex, for example a cluster of Si atoms such as the {311} linear defects often encountered during the early stages of Si RTP, must either diffuse as part of the defect complex or first disassociate from the complex so as to move as a free interstitial. Because interstitial mobility is inversely related to size, small atoms or clusters have a significantly higher mobility, at a fixed temperature, than do large ones. As a result, the transport of boron atoms as part of an atomic complex is small when compared to the transport of free boron

interstitials. Furthermore, the energy required for a boron atom to disassociate from a complex reduces the statistical likelihood that these "trapped" atoms will move throughout the silicon lattice. Finally, boron atoms that sit in substitutional sites – the activated boron – diffuse by a vacancy exchange mechanism. Vacancy exchange requires that a silicon vacancy move into close proximity with a substitutional boron atom so that the boron atom can jump from its current site into the nearby vacancy. This process is statistically unlikely, compared to the probability of a boron atom moving amongst interstitial sites. The other option is that the substitutional boron atom deactivate into a silicon site and move as a free interstitial. For this to occur, the boron atom must overcome an energy threshold for deactivation, reducing the probability of occurrence relative to a boron interstitial that is already free to diffuse. Both processes result in substitutional diffusion that is orders of magnitude slower than interstitial diffusion; therefore, active boron atoms are considered immobile relative to free interstitials. This relative immobility gives rise to the concentration enhanced diffusion discussed in Sec. 5.2.

The transport, J , of any species N through a material is governed by Fick-Einstein diffusion [1,48]:

$$J_N = L_{NN} \frac{d\mu_N}{dx} \quad 5-1$$

L_{NN} represents the transport coefficient for species N , μ_N represents the chemical potential and x is a generic distance variable. The spatial derivative of the chemical potential represents the chemical driving force for diffusion. To simplify the analysis and because it is prohibitively difficult, given today's level of diagnostic capabilities, to empirically measure diffusion in multiple dimensions, eqn. 5-1 is written in one dimension. Generalizing eqn. 5-1 to three dimensions is straight-forward but does not change the results.

By definition, the chemical potential of species N is written [49]:

$$\mu = \mu_N^o + RT \ln[a_N] \quad 5-2a$$

expanding the activity, a_N :

$$\mu = \mu_N^o + RT \ln[\gamma_N C_N] \quad 5-2b$$

The spatial derivative of eqn. 5-2 is:

$$d\mu/dx = d\mu_N^o/dx + RT \left(d\ln \gamma_N/dx + d\ln C_N/dx \right) \quad 5-3$$

Because μ_N^o and γ_N are constants, this reduces to:

$$d\mu/dx = RT d\ln C_N/dx \quad 5-3$$

Recognizing the mathematical integral identity:

$$d\ln X/dx = \frac{1}{X} dX/dx \quad 5-4$$

eqn. 5-3 evolves to:

$$d\mu/dx = \frac{RT}{C_N} dC_N/dx \quad 5-3$$

The transport coefficient, L_{NN} , for species N is composed of the mobility and concentration of species N. Writing L_{NN} to reflect this [48]:

$$L_{NN} = \frac{D_N C_N}{RT} \quad 5-5$$

For a dilute solution, the interaction between two atoms of species N is small and the second N, which represents these interactions, can be dropped. A very high dose, low energy boron implant has approximately 10^{20} B/cm³. Compared to the Si atomic concentration of $5 \cdot 10^{22}$

Si/cm³, a ratio of 500:1, the B atoms typically interact with Si atoms in the lattice are rarely come into contact with other B atoms. Equation 5-5 becomes:

$$L_N = \frac{D_N C_N}{RT} \quad 5-5$$

By Einstein's relationship, the mobility of species N is D_N/RT . Consequently, the mobility of a given species directly affects the diffusivity of that species.

Combining the transport coefficient and the chemical driving force yields a more familiar version of eqn. 5-1:

$$J_N(x) = -\frac{D_N C_N}{RT} \frac{dC_N(x)}{dx} \quad 5-6$$

The simplified result is the common form of Fick's first law:

$$J_N(x) = -D_N \frac{dC_N(x)}{dx} \quad 5-6$$

Whether transport is analyzed using eqn. 5-1 or eqn. 5-6, it is clear that both the mobility, through the diffusivity, and the local concentration drive the transport of any given species.

These fluid transport equations will be used to represent an average, or net, flow of boron atoms through the silicon. In a perfect crystal, boron atoms move in every direction without prejudice. When a concentration gradient exists, this random motion appears directed since a larger number of atoms are available to randomly move out of areas of high concentration and into areas of low concentration. The result is the net transport of atoms down a concentration gradient.

The movement of individual atoms, however, is not random. The activity of an atom depends upon its energy and the atoms or defects adjacent to it at any given time. For example, assume a concentration gradient exists in which the atoms in the areas of low

concentration are highly energetic and the atoms in the areas of high concentration have exactly zero energy. Atoms with zero energy remain stationary while highly energetic atoms move freely. In this scenario, a net diffusion of energetic atoms from the area of low concentration and into the area of high concentration occurs in apparent defiance to eqn. 5-6. Another, more common situation, involves the presence of defects. Defects trap atoms and create clusters as atoms randomly move into the region. The trapped atoms are caught in a potential well and cannot move free of the cluster until their energy exceeds the energy threshold of the potential well. This process is illustrated in Fig. 5-1. When enough energy is applied to the trapped atoms, typically by heating the crystal, the cluster dissolves as the atoms are now free to move through the lattice. Until this energy is applied, however, the local concentration of atoms in these clusters can grow to several orders of magnitude higher than the concentration of atoms several atomic spacings away.

Because thermal energy typically dominates during high temperature processing, and because this energy is transferred efficiently between atoms through numerous phonon collisions, the distribution of energy between atoms maintains a predictably gaussian distribution. As a result, the location of high and low energy atoms is random and the existence of

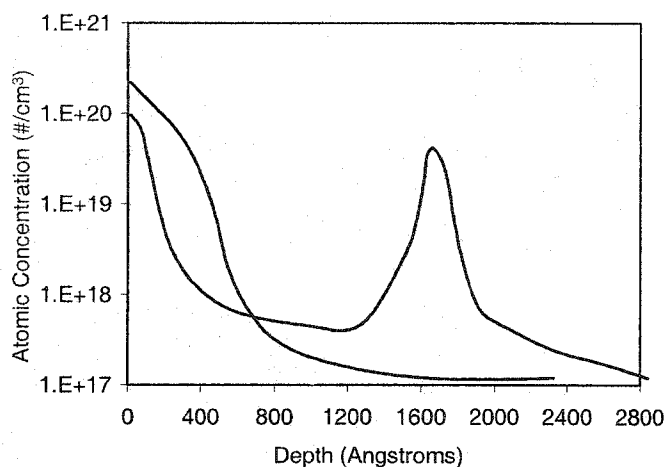


Fig. 5-1. An illustrative concentration profile showing how defects can trap atoms and lead to local concentrations that are much higher than anticipated by eqn. 5-1. In this cartoon, defects at 1600 Å have created a concentration peak.

a significant energy gradient could happen only in very unique circumstances. The result is a system in which predictable, energy driven atomic reactions occur. The random distribution of energy gives the cumulative action of these individual reactions a random appearance when viewed on a macroscopic scale. As a result, diffusion down a concentration gradient appears as atoms moving in a single direction. In fact, the atomic flux is still the result of numerous random motions. These numerous random motions are themselves the result of atomic interactions, driven specifically by the relative energy of each atom in the lattice.

While the energy distribution of the atoms may be sufficiently random, the location of lattice defects, including substitutional or interstitial impurities, that can react with either boron atoms or silicon defects is not. To account for the realization that atoms react differently depending on their relative location in the lattice and the number and identity of adjacent atoms, the transport, diffusivity and concentration terms in eqn. 5-1 and 5-6 must be generalized to account for each possibility as shown in eqn. 5-7:

$$J_N(x) = -\sum_N D_N \frac{dC_N(x)}{dx} \quad 5-7a$$

$$J_N(x) = -\sum_N L_N X_N \quad 5-7b$$

In eqn. 5-7b, the driving force for diffusion is replaced by the general term, X_N . Using this identification scheme, the diffusivity and concentration of a given species, in each possible scenario, are accounted for. In accordance with the random distribution of energy, the diffusivity is represented by:

$$D = D_o \exp(-Q/kT) \quad 5-8$$

Q is the specific enthalpy for diffusion to occur, D_o is a standard pre-exponential coefficient, and kT is the thermal energy.

Accounting for every individual interaction in this fashion is, of course, unreasonable in practice. As such, an average, or effective, diffusivity will be assigned for each general scenario considered. Applying this understanding of interstitial mobility facilitates a refined picture of boron diffusion during the first few seconds of the anneal. Eqn. 5-7 is re-written:

$$J_B(x) = -D_{B,S} \frac{dC_{B,S}(x)}{dx} - D_{B,I} \frac{dC_{B,I}(x)}{dx} \quad 5-9$$

to account for the varying mobility between interstitial and substitutional boron.

$D_{B,S}$ and $D_{B,I}$ are the diffusivity of substitutional and interstitial boron respectively and $C_{B,S}$ and $C_{B,I}$ are the concentrations of substitutional and interstitial boron. Boron moves onto substitutional sites by the defect reaction:



and applying the law of mass action yields:

$$B_{Si} = \sqrt{V_{Si} \cdot B_I} \exp\left(\frac{-E_A}{kT}\right) \quad 5-11$$

In eqns. 5-10 and 5-11, B_I , V_{Si} , B_{Si} , and h represent boron interstitials, silicon vacancies, boron atoms sitting in substitutional sites, and holes respectively. E_A is the energy required for the reaction to occur, 4.9 eV.

Interstitial boron atoms, which are completely free from associations with other point or extended defects, have a mobility, and therefore diffusivity, D_I , that is several orders of magnitude greater than that of substitutional boron. As a result, the second half of eqn. 5-9 dominates boron transport whenever the $[B_I]$ is on the order of $[B_{Si}]$. The first half of eqn. 5-9 only plays a role in boron transport when the $[B_I]$ is very small relative to the $[B_{Si}]$. Factors leading to the creation of interstitial boron atoms or diminishing the ability of other impurities to bond with boron interstitials (i.e. to form a defect complex) enhance the

transport of boron. For example, increasing the dose of implanted boron leads to a moderately higher $[B_{Si}]$ and a dramatically higher $[B_I]$. The result is greater boron diffusion and a deeper junction for a given thermal treatment [50,51]. Furthermore, because B_{Si} atoms are relatively immobile, any reaction that displaces boron from substitutional sites enhances the average mobility of the boron atoms. Conversely, a scenario that encourages boron atoms to reside in silicon sites lowers the total boron transport, eqn. 5-9, reducing diffusion. Finally, the introduction of secondary impurities, which interact with free B_I to form bonded complexes, reduce the effective concentration of free B_I in favor of these less mobile complexes. This requires additional terms for eqn. 5-7, which will be presented and discussed in Sec. 5-4.

5.2 Concentration Enhanced Diffusion

Because of the nature of high-dose, low energy ion implantation, concentration enhanced diffusion (CED) plays a dominant role in determining the final dopant profile. CED is a by-product of the relatively immobile nature of substitutional boron, which must overcome an energy barrier associated with detaching itself from the silicon site in which it resides before it is free to diffuse. As discussed, interstitial boron sees no such energy barrier and therefore diffuses much more rapidly. When the concentration of boron is low, relative to the number of available substitutional silicon sites, most boron atoms are (or quickly become) substitutional, and therefore less mobile than interstitial boron atoms. On the other hand, when the concentration of boron is super-saturated, relative to the number of available silicon sites, most boron atoms remain as interstitials and therefore are very mobile. The result is an apparent energy barrier for diffusion that decreases as the concentration of boron increases.

After a high dose ($>10^{15}/\text{cm}^2$) boron implant, most of the implanted atoms within the top 20 – 30 nm of the surface are interstitials. Most of the boron atoms at the end of the junction (concentration $< 10^{18}/\text{cm}^3$) are substitutional. Utilizing eqn. 5-9, the interstitial component dominates near the surface while the much slower, substitutional component dominates when

the boron concentration falls below $10^{18}/\text{cm}^3$. In other words, the average mobility of boron increases with concentration, and the transport of boron in the near surface of the silicon is disproportionately higher than that of boron located only a few tens of nanometers deeper in the silicon. This is illustrated in Fig. 5-2. As the highly mobile interstitial atoms move from the super-saturated surface deeper into the silicon, their local concentration diminishes, they more readily find substitutional sites, and the fast interstitial diffusion gives way to slower substitutional diffusion. Because the transport of boron varies so dramatically as a function of depth in silicon, a rapid diffusion/deceleration process occurs, causing a saturation in the concentration of boron in the region between the surface and a specific interstitial-substitutional concentration "threshold". The result is an extremely "box-shaped", or abrupt, dopant profile as compared to the Gaussian distribution seen during traditional Fickian diffusion of an impurity species.

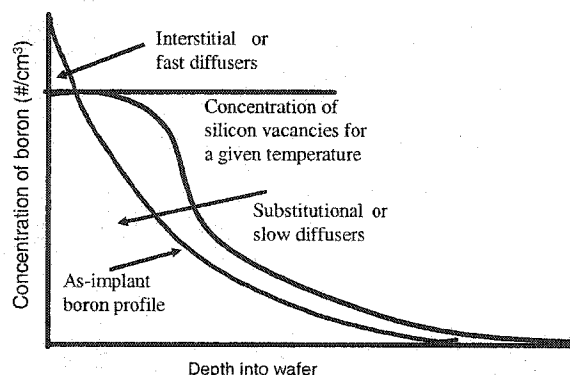


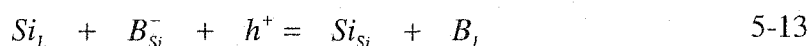
Fig. 5-2. An example of concentration enhanced diffusion. The fast interstitials quickly catch up to the slower substitutional atoms. This mobility gradient results in a "box-shaped" profile.

5.3 Maximizing Activation, Minimizing Diffusion

The energetics for boron activation in an undamaged silicon lattice, eqn. 5-5, are extremely unfavorable. As a result, achieving high levels ($>10^{20}/\text{cm}^3$) of boron activation is nearly impossible in undamaged silicon. Fortunately, the implantation process badly



damages the silicon, displacing a significant fraction of the silicon atoms as interstitials [50,52]. While the V_{Si} are distributed throughout the damaged region, the nature of ion implantation, which bombards the surface with high energy ions, forces the Si_I to accumulate at the end of the damaged region. Upon heating to $\sim 400 - 600^\circ\text{C}$, these Si_I cluster together in line defects along the $\{311\}$ axis in the end-of-range damage region, leaving the V_{Si} available for boron activation in the early stages of the anneal. Above 700°C , these $\{311\}$ defect clusters begin to dissolve at an appreciable rate, and the emitted Si_I are free to compete with B for silicon sites. This competition forces a significant fraction of the activated boron atoms out of substitutional sites and into interstitial sites. Forced into interstitial status, the



B_I are free to move rapidly throughout the silicon. The result is an *apparent* enhanced diffusion as these $\{311\}$ defects dissolve. In reality, a greater fraction of boron transport is now governed by the fast interstitial mechanism of eqn. 5-9 as opposed to the slower substitutional mechanism. This "enhanced" diffusion is seen consistently, as reported in the literature [8-12,13,36,37,39], during the time these $\{311\}$ clusters dissolve. Before these clusters dissolve, no enhanced diffusion is seen. Once the $\{311\}$ cluster dissolution is complete, this Si "kick-out" mechanism and the perceived enhanced diffusion of B subsides.

The logical way to avoid this sort of de-activation and rapid diffusion is to prevent the dissolution of these {311} Si_I defect clusters. Because these defects begin dissolving at an appreciable rate above 700°C , annealing below this temperature prevents the release of Si_I , thereby eliminating the competition for V_Si illustrated by eqn. 5-13. The expected result is a higher concentration of active boron with no net diffusion of the implanted boron. Experimental results with this process, termed solid phase epitaxy, SPE, show no net diffusion when annealing is performed below 700°C [53,54]. These results also show levels of activation (see Fig. 5-23 in Sec. 5.7) that are superior to those achieved with high temperature spike anneals. Unfortunately, the damage created by the implant process is not effectively repaired at low temperatures and the final devices are inoperable. In addition, future high temperature processing results in continued diffusion and deactivation if the {311} defects are allowed to dissolve. For high quality devices, a high temperature anneal step is needed to repair the implant damage and to eliminate the {311} Si_I line defects.

The next best solution to a low temperature anneal is a high temperature anneal with a minimal amount of time spent above 700°C . This process, known as spike annealing, activates the boron, repairs the damage, and leaves as little time as possible for the released Si_I to displace the active boron atoms from substitutional sites. In microwave spike annealing, the wafer spends only a few seconds above 700°C . The excellent junctions formed by microwave spike annealing (see Figs. 5-4 – 5-12 in Sec. 5.4) demonstrate the effectiveness of this technique. It has been shown that continued increases in ramp rate result in less time spent above 700°C and less diffusion during the anneal [55]. Taking this minimal time above 700°C to the extreme, researchers at Vortek performed lamp-based spike anneals to 1300°C with only a few milliseconds above 700°C [55,56]. The $\text{R}_\text{s}\text{-X}_\text{j}$ results,

some of which are shown in Fig. 5-23, indicate a no net diffusion anneal process with activation superior to that achieved with conventional spike anneal technologies. While the {311} defects did dissolve, deactivation and subsequent rapid interstitial diffusion were kinetically limited by minimizing the time during which the B_i and Si_i atoms had sufficient thermal energy to move. These experiments also show that B activates in large quantities during the first few milliseconds of the anneal, which is consistent with the super-saturation of V_{Si} before the dissolution of the {311} defects.

5.4 Effect of Environmental Oxygen on Boron Diffusion and Activation

Oxygen, being extremely difficult to remove, is the most common impurity in silicon. Lamp-based RTP studies [32] have shown that an oxidizing ambient both enhances the diffusion, see Fig. 5-3, and hinders the activation of any species of implanted boron (i.e. B^+ , BF_2 , pulsed plasma doped BF_3). Maintaining a low background oxygen concentration, $[O_2]$, of 100 ppm during the spike anneal has broken the previous stalemate in anneal technology, allowing the 90 nm node to be achieved for the first time using lamp-based RTP [23].

The effect of $[O_2]$ in the ambient during microwave annealing was studied by running several anneals at various $[O_2]$ for B^+ , BF_2 , and pulsed plasma-doped BF_3 implants. Fig. 5-4 shows the boron concentration profiles for a 4500 eV BF_2 implant microwave spike annealed to 1050°C. Fig. 5-5 shows the boron concentration profiles for a 2200 eV BF_2 implant microwave spike annealed to 950°C. Fig. 5-6 shows the boron concentration profiles for a 2200 eV BF_2 implant microwave spike annealed to 1000°C. Fig. 5-7 shows the boron concentration profiles for a 500 eV BF_2 implant microwave spike annealed to various final temperatures. Fig. 5-8 shows the sheet resistance and junction depth (measured at $10^{18}/cm^3$)

results for a 4500 eV BF_2 implant microwave spike annealed to 1000°C . Fig. 5-9 shows the boron concentration profiles for a 800 eV BF_3 implant microwave spike annealed to 950°C . Fig. 5-10 shows the boron concentration profiles for a 2200 eV BF_3 implant microwave spike annealed to 990°C . Fig. 5-11 shows the boron concentration profiles for 250 eV B^+ and 500 eV B^+ implants microwave spike annealed to 950°C . Fig. 5-12 shows the boron concentration profiles for 250 eV B^+ and 500 eV B^+ implants microwave spike annealed to 1050°C .

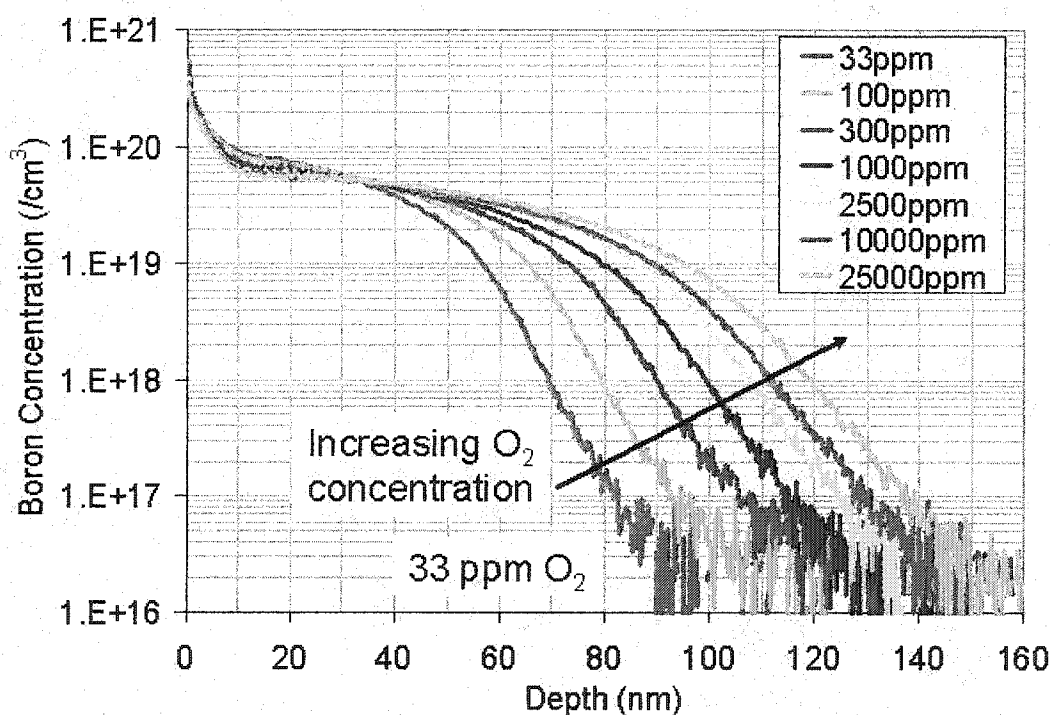


Fig. 5-3. Oxygen enhanced diffusion effects in lamp-based annealing of ^{11}B implants. Anneal step was 10 seconds at 1000°C . Figure was reprinted from 32 with permission.

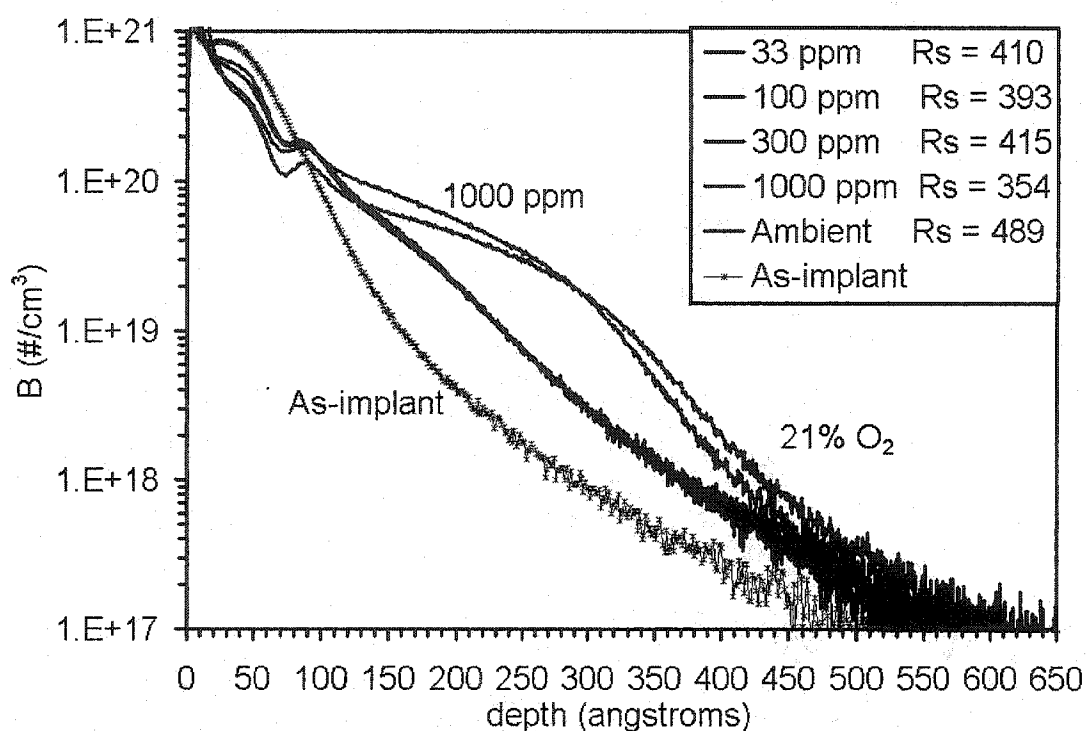


Fig. 5-4. 4500 eV BF_2 implant, $10^{15}/\text{cm}^2$ microwave spike annealed to 1050°C

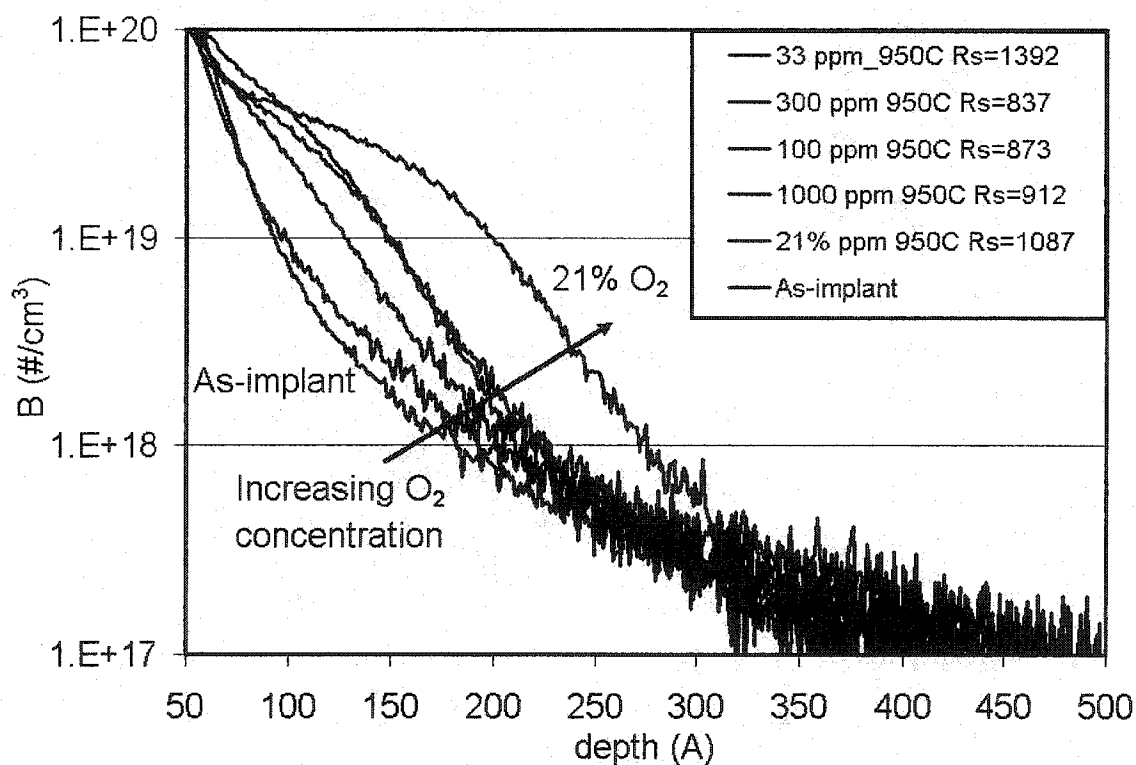


Fig. 5-5. 2200 eV BF_2 implant, $10^{15}/\text{cm}^2$ microwave spike annealed to 950°C

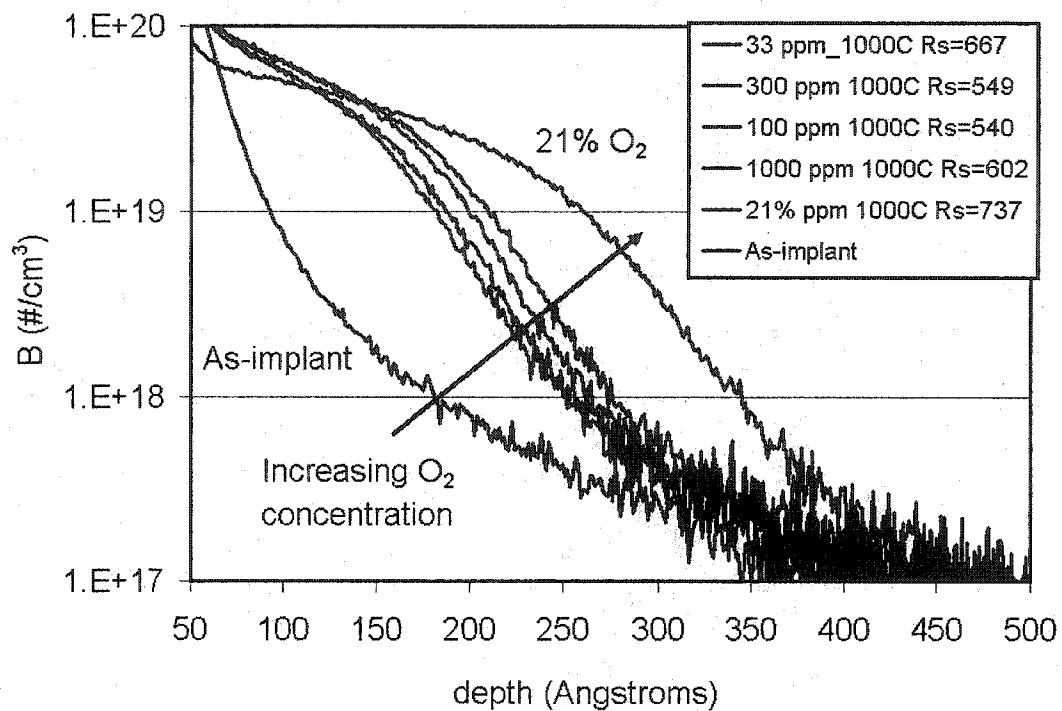


Fig. 5-6. 2200 eV BF_2 implant, $10^{15}/\text{cm}^2$ microwave spike annealed to 1000°C

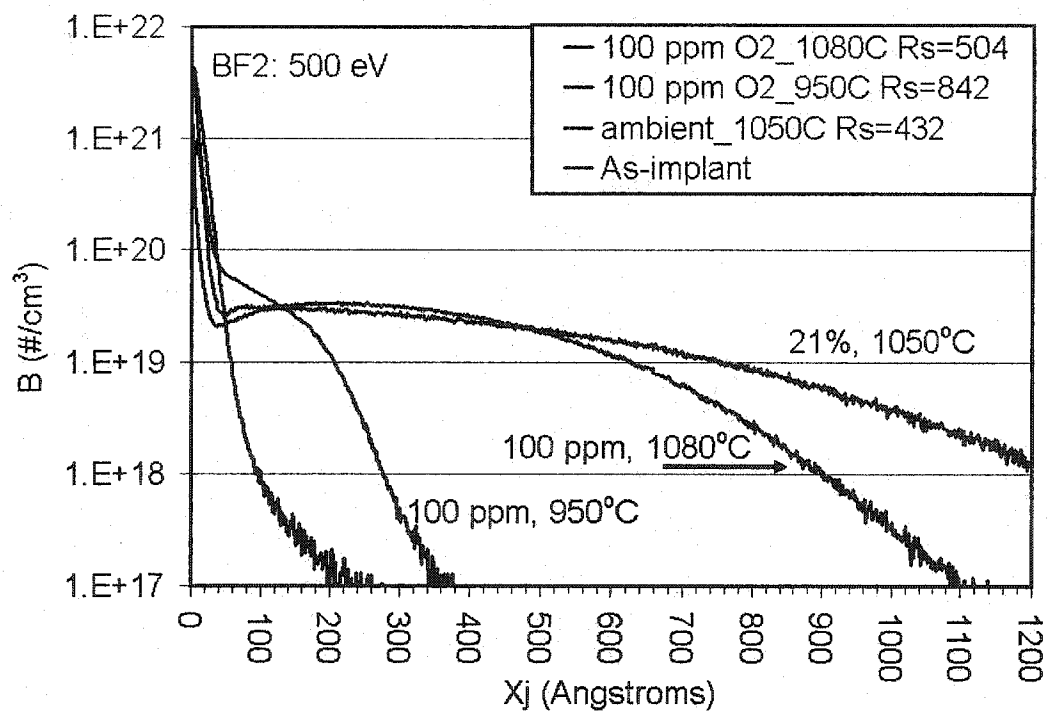


Fig. 5-7. 500 eV BF_2 implant, $10^{15}/\text{cm}^2$ microwave spike annealed.

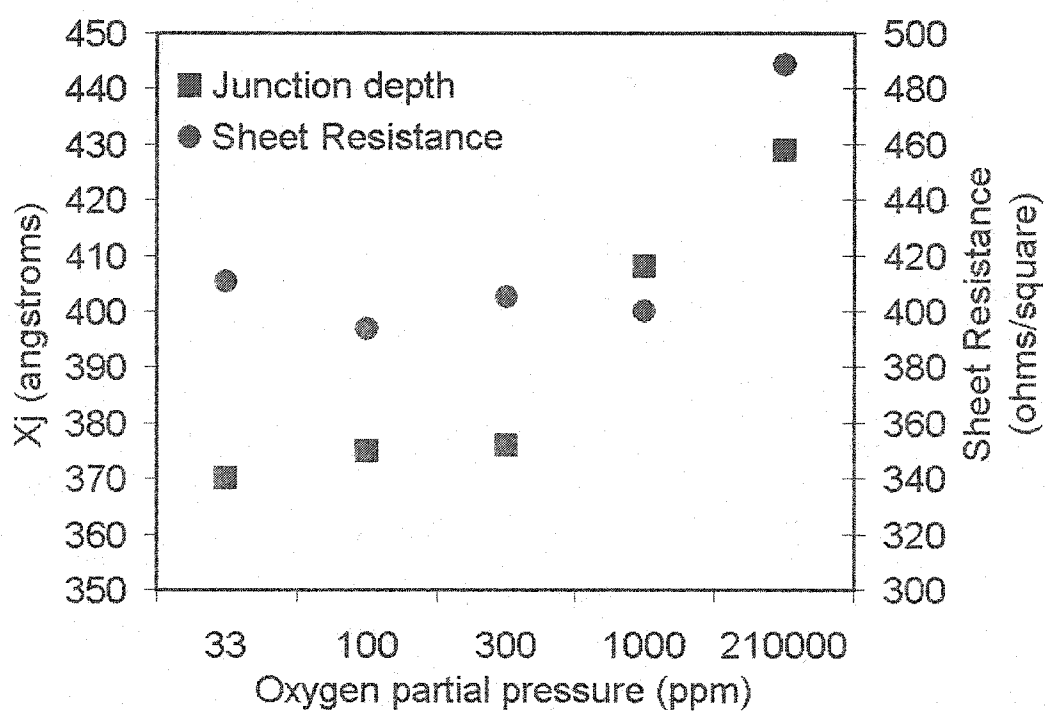


Fig. 5-8. 4500 eV BF₂ implant, 10¹⁵/cm² microwave spike annealed to 1000°C.

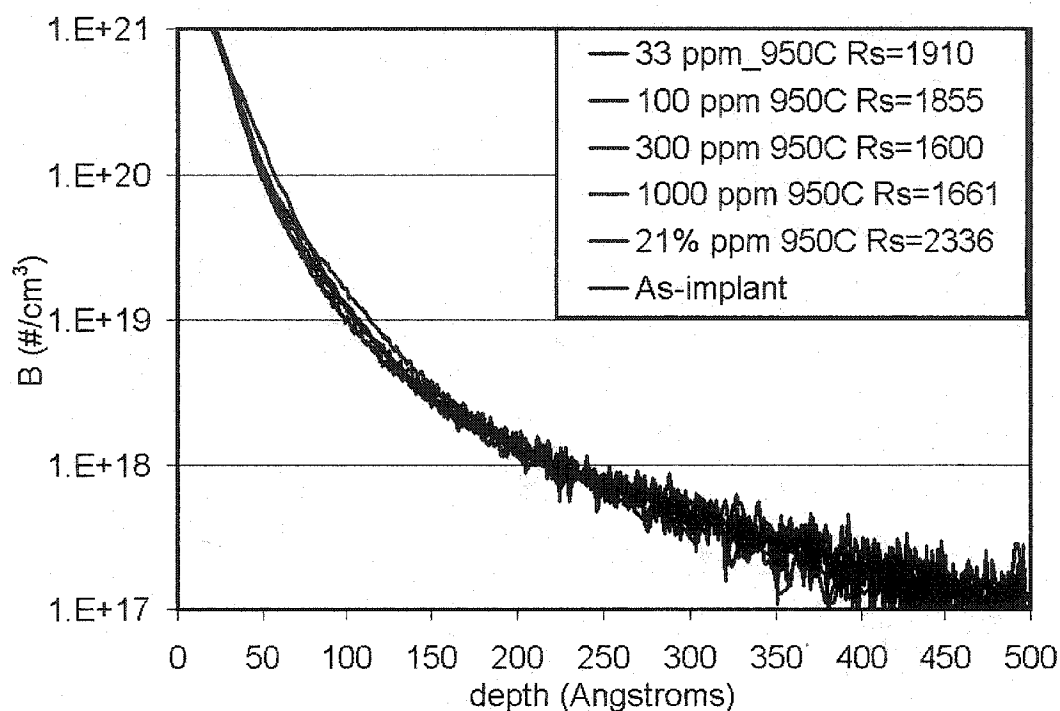


Fig. 5-9. 800 V BF_3 implant, $5 \times 10^{15}/\text{cm}^2$ microwave spike annealed to 950°C

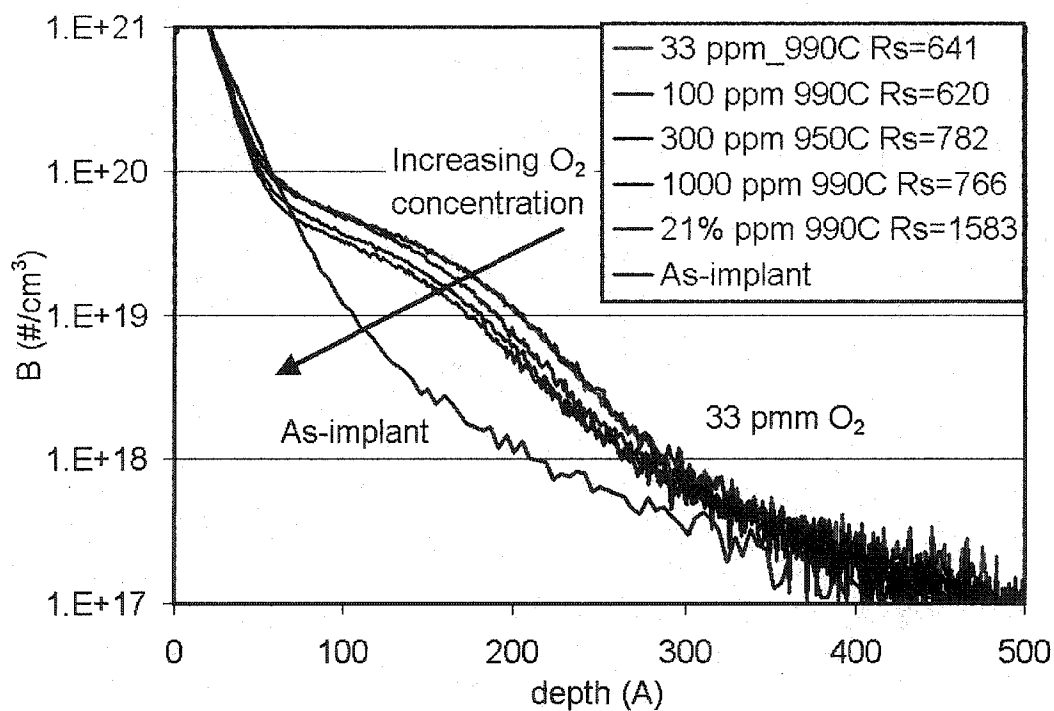


Fig. 5-10. 800 V BF_3 implant, $5 \times 10^{15}/\text{cm}^2$ microwave spike annealed to 990°C

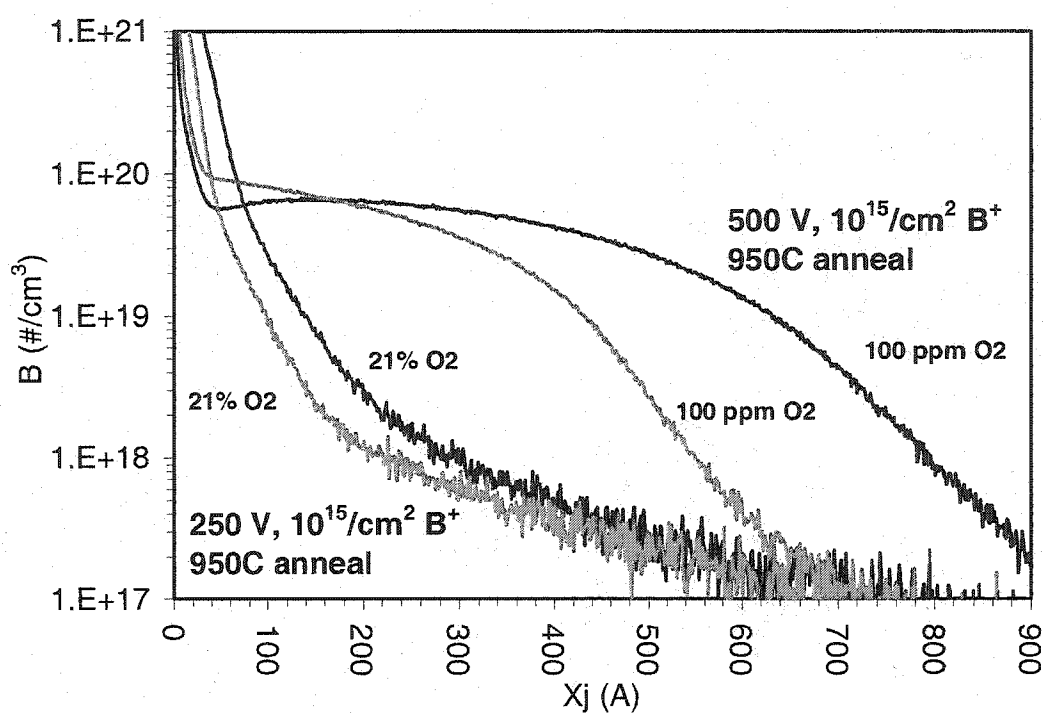


Fig. 5-11. 250 eV and 500 eV $10^{15}/\text{cm}^2$, B implants microwave spike annealed to 950°C.

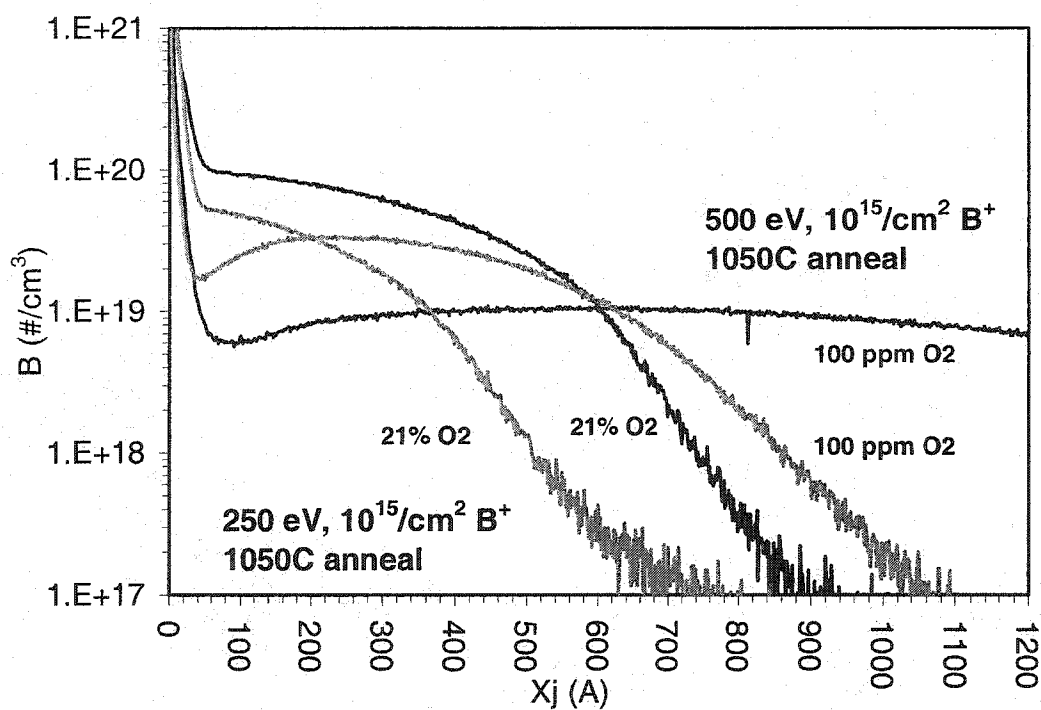


Fig. 5-12. 250 eV and 500 eV $10^{15}/\text{cm}^2$, B implants microwave spike annealed to 1050°C

The BF_2 results, Figs. 5-4 – 5-8, show diffusion and activation trends that are very similar to lamp-based RTP. The boron concentration profiles for each BF_2 implant/anneal scenario (Fig. 5-4 – 5-7) indicate that the diffusion of boron increases with increasing $[\text{O}_2]$. This was seen in previous lamp-based anneal studies (Fig. 5-3), and it has been classified as "oxygen enhanced diffusion," OED. Although the mechanisms that drive OED are still in debate, as the $[\text{O}_2]$ is reduced, the kinetics of oxidation are similarly reduced, and the perceived oxygen enhanced diffusion effects are diminished. In addition, the sheet resistance decreases with increasing $[\text{O}_2]$ until coming to a minimum value between 100 - 1000 ppm O_2 . After this minimum, the sheet resistance increases with further reduction in $[\text{O}_2]$. This trend is distinctly illustrated by plotting the sheet resistance and junction depth (measured at $10^{18}/\text{cm}^3$) as a function of $[\text{O}_2]$, Fig. 5-8. This minimum in sheet resistance represents the point at which improvements in dopant activation, and subsequently lower silicon resistivity, are maximized with respect to $[\text{O}_2]$. Any further reduction in $[\text{O}_2]$ beyond this point results in less diffusion and a shallower junction. As discussed in Ch. 2, reducing the junction depth without improving dopant activation leads to an increase in sheet resistance without changing the quality of the junction. This is precisely what is seen in the sheet resistance data from Fig. 5-4–5-8.

Boron concentration profiles for the microwave annealed pulsed plasma BF_3 implants, Fig. 5-9 and 5-10, indicate a different diffusion trend with $[\text{O}_2]$. Very little difference in the net diffusion of boron is seen for these implants as the oxygen concentration is varied. This suggests that the diffusion of boron, when implanted by pulsed plasma doping, has very little sensitivity to the $[\text{O}_2]$ in the ambient during microwave spike annealing. The sheet resistance

does, however, fall from 1583 ohms/square in an uncontrolled (21% oxygen) ambient to 620 ohms/square at an $[O_2]$ of 100 ppm during annealing to 990°C.

There are two key differences between the pulsed plasma BF_3 implants and the traditional beamline BF_2 implants that may explain the different trends in boron diffusion. First, beamline implantation accelerates the dopants toward the wafer at a very specific energy. This results in a Gaussian concentration profile after the implantation process. Plasma doping, on the other hand, pulses the wafer at a target implant energy, but invariably samples the entire range of energies below this energy target. Consequently, the pulsed plasma doped boron concentration profile is retrograde instead of Gaussian. Second, beamline implantation of BF_2 yields a 2:1 F:B atomic ratio in the silicon. With pulsed plasma doping, this ratio falls to 1:1. The result is less F in the wafer after P²LAD implantation and during the subsequent anneal. Because lamp-based RTP studies show no difference in the oxygen enhanced diffusion trends between these two doping technologies [32], the existence of a different OED trend during microwave annealing indicates an apparent difference between lamp-based and microwave anneal technologies. This difference is discussed in detail in section 5.5.

The boron concentration profiles obtained for microwave annealing B^+ implants, Fig. 5-11 and 5-12, show the exact opposite trend as that seen during both lamp-based annealing of B^+ , Fig. 5-2, and microwave annealing of BF_2 , Fig. 5-4 – 5-8. A marked *increase* in diffusion as the $[O_2]$ is restricted is seen for both the 250 eV and the 500 eV implants. The sheet resistance of the low $[O_2]$ microwave annealed B^+ implant samples was also higher than that of the uncontrolled case. Specifically, the R_s was 279 ohms/square when 500 eV B^+ implanted samples were microwave annealed in the uncontrolled, ~21% oxygen ambient compared to 508 ohms/square when microwave annealed in the 100 ppm $[O_2]$ ambient. This

trend of increased diffusion (higher mobility boron) and higher Rs (less activation) is consistent with an increased amount of implanted boron remaining in interstitial sites after microwave annealing is performed in a low [O₂] ambient.

The dramatic differences witnessed when annealing in an uncontrolled (~21% O₂ in N₂) versus a controlled low ppm O₂ ambient, Figs. 5-4 – 5-10, are partially attributed to the oxidation of the top 30-40 Å of silicon. After implantation, the largest fraction of the implanted dose lies in the extreme near surface, i.e. the top 50 Å, of silicon. The volumetric concentration of boron (#/cm³) greatly exceeds the available substitutional sites, and the majority of these boron atoms sit in interstitial sites. For superior junctions to form, these dopant atoms must diffuse into the near-surface regions until they find an available substitutional site and activate. When annealing in an oxidizing ambient, it is foreseeable that this oxidation process, which consumes equal amounts of silicon and molecular oxygen, will either oxidize large quantities of the implanted boron, forming B₂O₃, or trap large quantities of boron in the oxide layer. This is especially true for impurities like boron, which, unlike phosphorous, are excellent network formers in glasses [33]. Because the B is super-saturated in the Si surface, there is no driving force for B diffusion from the oxide into the Si. It is unlikely, therefore, that the trapped or oxidized boron diffuses back into the silicon on the time scales involved in a spike anneal.

Careful inspection of the SIMS plots in Figs. 5-4 – 5-10 indicates that the incorporation of boron in the oxide approaches 10% of the total implanted B dose, and the surface concentration of B approaches 10²¹/cm³, easily exceeding 1% of the silicon concentration. The oxide thickness is difficult to gauge since ellipsometric measurement of very thin dielectric films is highly inaccurate when the chemical composition is not known precisely.

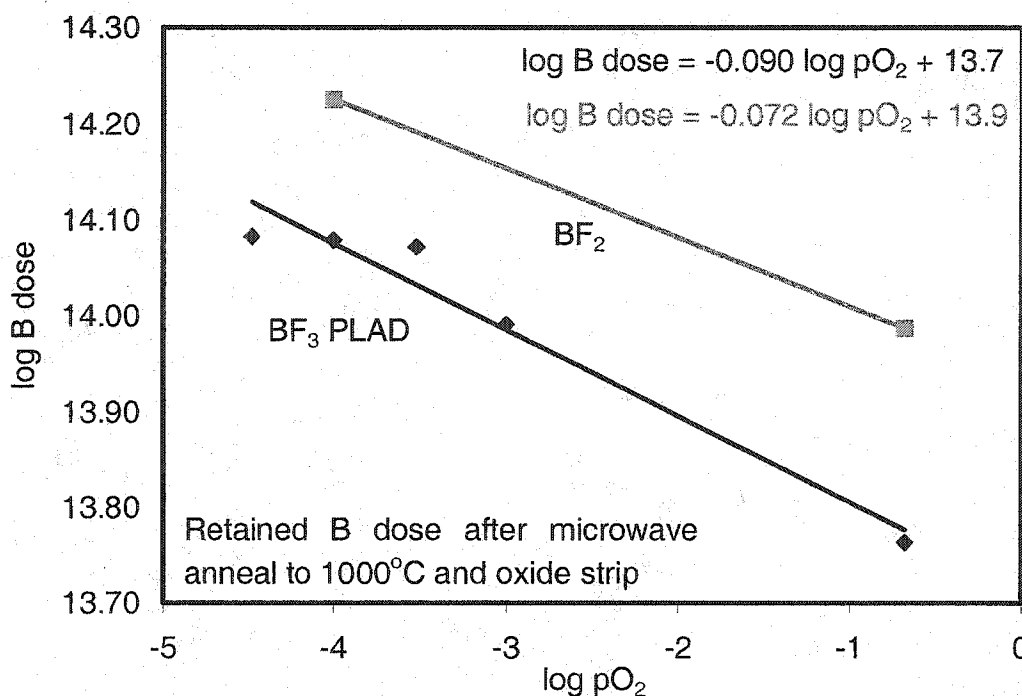


Fig. 5-13. Log-Log plot of B retained as a function of oxygen concentration, [O₂], after a microwave spike anneal. The BF₂ samples are from Fig. 5-5, and the BF₃ samples are from Fig. 5-9.

Nevertheless, the amount of boron dose trapped in the oxide can be indirectly measured by stripping the grown oxide layer before SIMS analysis and calculating the retained dose. The dose retained (total boron dose, *not* active boron) in the silicon after microwave spike anneals for 2200 eV BF₂ and 800 eV BF₃ P²LAD, Fig. 5-13, illustrate this trend. At high [O₂], the retained dose falls considerably. At lower [O₂], the retained dose increases until saturating at 300 ppm. Below 300 ppm O₂, the oxidation kinetics are effectively curtailed for the spike anneal time scales used in these experiments. These results clearly point to an increased consumption of dose when the anneal process is performed in an oxidizing ambient.

A summary of the amount of boron dose retained in the silicon after several of the anneals is tabulated in Table 1. Despite having a similar sheet resistance at low [O₂], 641

Table 1.

Amount of boron dose retained in the silicon for 2200 eV, $10^{15}/\text{cm}^2$ BF_2 implants and 800 V, $10^{15}/\text{cm}^2$ BF_3 P²LAD implants after microwave spike annealing in a variety of $[\text{O}_2]$. The error range for the SIMS analysis is 5%. The final microwave spike anneal temperature was 1000°C for the 2200 eV samples and 990°C for the 800 V samples.

Implant species and energy	$[\text{O}_2]$ in the anneal ambient	Retained B dose
2200 V BF_2	100 ppm	1.68×10^{14}
2200 V BF_2	210000 ppm	9.55×10^{13}
800 V BF_3	33 ppm	1.26×10^{14}
800 V BF_3	100 ppm	1.20×10^{14}
800 V BF_3	300 ppm	1.17×10^{14}
800 V BF_3	1000 ppm	9.77×10^{13}
800 V BF_3	210000 ppm	5.75×10^{13}

ohms/square for plasma doping and 667 ohm/square for beamline implantation, the plasma doped samples have a lower overall retained dose than the beamline BF_2 samples. The percentage of active dose is higher for the P²LAD samples than the beamline implants. Although a large quantity of dose is consumed at high $[\text{O}_2]$ for both implant sets, the large percentage of "wasted" or inactive dose for the beamline implants insulates its final sheet resistance from the oxidation effects discussed. Furthermore, the retrograde boron concentration profile, obtained during plasma doping, leaves a higher percentage of the overall dose nearer to the silicon surface. As a result, the consumption of the near surface silicon, by an oxide layer, consumes a greater fraction of the implanted dose from the BF_3 samples than from the BF_2 samples. This is exactly what is seen in Table 1 and Fig. 5-13.

Maintaining a low $[O_2]$ of ~ 100 ppm in the ambient during the anneal limits the oxide growth. This, in turn, limits the boron dose loss, leaving the boron free to diffuse into the silicon. When an appropriate thermal process is applied, the free boron diffuses from the super-saturated near-surface region into the next 20 nm of silicon without moving deep into the wafer. For example, in Fig. 5-7 the 1050°C microwave spike anneal in an uncontrolled ambient results in a junction depth of 120 nm. Lowering the $[O_2]$ results in a junction depth of 110 nm at the *higher* anneal temperature of 1080°C . Reducing the microwave anneal temperature to 950°C and controlling the $[O_2]$ to ~ 100 ppm results in a junction depth of only 27 nm. The other microwave annealed BF_2 implants, Fig. 5-4 – 5-7, show similar trends. Finally, plotting the junction depth and sheet resistance data as a function of $[O_2]$, Fig. 5-8, displays these results in a remarkably clear fashion.

The *perceived* concentration enhanced diffusion (CED) effect, described in Sec. 5.2, that occurs in highly-doped silicon layers facilitates this short range diffusion, resulting in the formation of a "box-shaped" boron concentration profile. By preventing excessive surface oxide growth, the driving force for interstitial diffusion, i.e. see eqn. 5-9, is free to dominate the kinetics of boron transport over incorporation into the surface oxide. This "box-shaped" profile is often quantified by the distance into the wafer it takes for the boron concentration to fall from $10^{19}/\text{cm}^3$ to $10^{18}/\text{cm}^3$. In Fig. 5-7, for example, the 1050°C microwave anneal in an uncontrolled ambient results in a profile that requires 45 nm to fall from $10^{19}/\text{cm}^3$ to $10^{18}/\text{cm}^3$. Restricting both the thermal budget and the oxygen concentration, as in the 950°C , 100 ppm $[O_2]$ microwave spike anneal, results in a profile that requires only 6 nm to fall from $10^{19}/\text{cm}^3$ to $10^{18}/\text{cm}^3$. It is clear from Fig. 5-4 – 5-8 that the overall result of ambient and thermal budget control during microwave spike annealing of BF_2 implanted samples is a

more "box-shaped", abrupt profile that incorporates a higher concentration of active dopants in the junction region.

The growth of a surface oxide provides an adequate explanation for the changes in R_s seen when annealing is performed in an oxidizing ambient. It cannot, however, explain the *apparent* enhancement of boron diffusion with increasing $[O_2]$ seen in Fig. 5-4 – 5-8. It also does not explain the reversal of this trend as seen in Fig. 5-9 – 5-12. An alternative theory is that the consumption of silicon vacancies by oxygen atoms inhibits the electrical activation of boron [34,35]. Approximately 40 years ago, vibrating reed spectroscopy was used to discover that oxygen resides in tetrahedral (i.e. the "substitutional") sites within the silicon lattice. The proposed defect structure, Fig. 5-14, achieves stability as oxygen donates two electrons to the silicon lattice. Instead of becoming electrically active, however, these two donated electrons complete

the eight electron outer shell, forcing the two remaining silicon atoms to bond covalently. The result is a silicon vacancy that is occupied, in a neutral fashion, by an oxygen atom.

As illustrated by eqn. 5-5 and 5-6, the activation of boron depends on both the

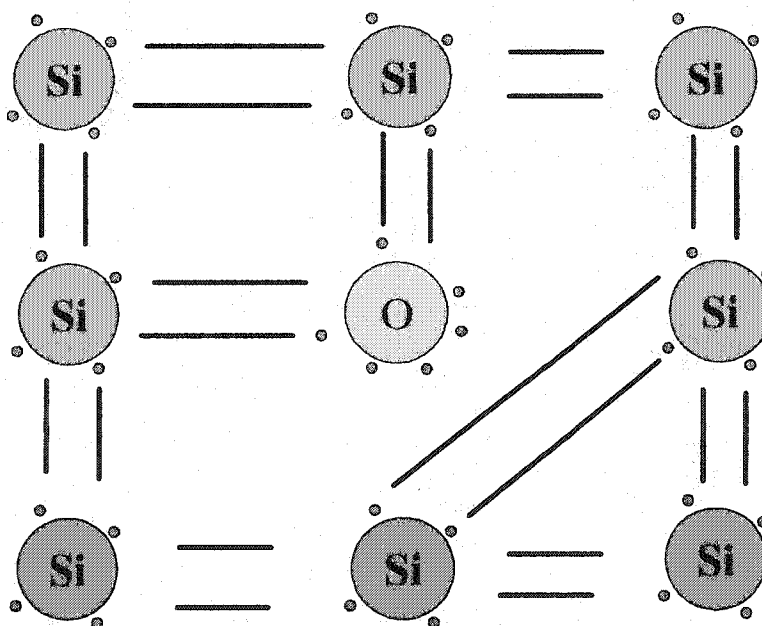


Fig. 5-14. Atomic interaction by which an oxygen atom consumes a substitutional site.

concentration of boron atoms and of silicon vacancies. A reduction in the number of silicon vacancies clearly reduces the probability of boron activation. The defect reaction for the consumption of V_{Si} by O atoms, as illustrated in Fig. 5-14, is written:



Applying mass action yields:

$$O_{Si} = V_{Si} \bullet O_I \exp\left(\frac{-E_o}{kT}\right) \quad 5-15$$

E_o represents the energy requirement for oxygen to consume a silicon site.

While the value of E_o is not known, Fig. 5-15 shows that the oxygen concentration after microwave annealing in an uncontrolled ambient (~21% oxygen) is an order of magnitude higher than when annealing is performed in a controlled 100 ppm $[O_2]$ ambient. Furthermore, the oxygen concentration in Fig. 5-15 is approximately two orders of magnitude greater than the boron concentration in the near surface region. Equation 5-15 indicates that the number of silicon vacancies consumed by oxygen atoms increases proportionately with $[O_2]$. Even if E_o is significantly larger than the threshold for boron activation, it is reasonable to conclude that, because of the

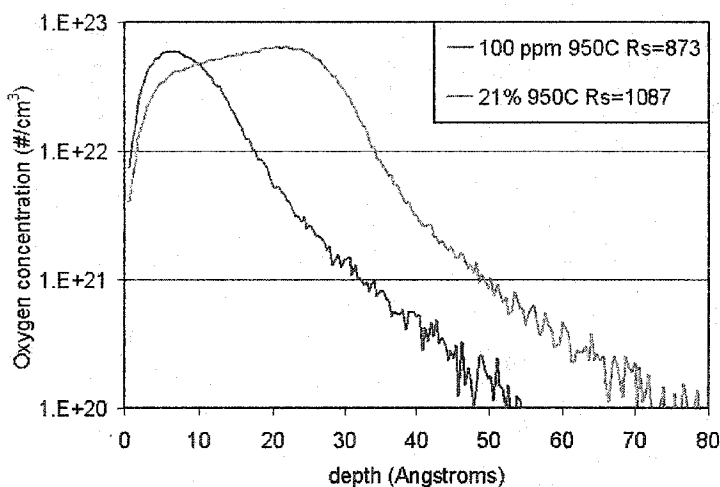


Fig. 5-15. Oxygen concentration for a 2200 eV BF_2 implant, $10^{15}/cm^2$ after a microwave spike anneal to 950°C (see Fig. 5-4) in an ambient whose oxygen concentration was controlled to 100 ppm and in an uncontrolled ambient.

large concentration of oxygen atoms, the statistical probability of an oxygen atom forming the defect complex shown in Fig. 5-14 [35,36] is large enough to significantly reduce the concentration of available V_{Si} . The result is a lower fraction of substitutional boron atoms when annealing is performed in an uncontrolled, high ppm O_2 ambient. The higher $[B_I]$, results in an increase in the magnitude of the interstitial component of eqn. 5-9. In other words, the magnitude of the rapid interstitial diffusion coefficient, which already dominates boron transport, increases. The result is a more rapid transport of boron in silicon.

Typically reported as oxygen "enhanced" diffusion, this phenomena is seen during both microwave annealing, Fig. 5-3 – 5-8, and lamp-based annealing as reported in the literature [32]. In addition to increasing boron transport, interstitial boron does not contribute to the concentration of electrical carriers. Therefore, this phenomena also partially accounts for the higher sheet resistance measurements seen in Fig. 5-3 – 5-10. While this trend is not consistent with microwave annealing of B^+ implants (Fig. 5-11 and 5-12), it is consistent for lamp-based annealing of B^+ implants and for the microwave annealing of BF_2 and P^2LAD BF_3 implants. This result suggests the impact of a microwave ponderomotive force on the equilibria setting O_{Si} and B_{Si} during annealing. Excluding this extra variable, the theory that oxygen atoms consume silicon vacancies, thereby leading to an enhanced boron diffusion and lower activation, is consistent with the experimental results in this work and with those reported elsewhere. Furthermore, this same oxygen "enhanced" diffusion effect is seen for phosphorous and arsenic, impurity atoms that, like boron, diffuse by an interstitial mechanism [58].

5.5 Effect of Implanted Fluorine on Boron Diffusion and Activation

Regardless of the heating technique employed, i.e. microwave, lamp-based, furnace, laser, etc., fluorine plays an important role in controlling the diffusion and activation of boron during high temperature processing [36-39]. It is well established that F interacts with both B and Si to create the B-F and Si-F defect associate pairs [36,37,38,39]. Downey et. al., for example, found that fluorine both hinders the diffusion and enhances the activation of boron in silicon. Robertson and Jones performed numerous experiments with fluorine co-implants, verifying and further expanding on these results [38,39]. A set of comparison anneals, Fig. 5-16, illustrates that the effect of fluorine on the diffusion and activation of boron during microwave annealing is the same as that seen during lamp-based annealing. In this figure, a 2200 eV, $10^{15}/\text{cm}^2$ BF_2 implant and a 500 eV, $10^{15}/\text{cm}^2$ B^+ implant are microwave spike annealed to 1050C. The BF_2 sample showed less boron diffusion with only a moderate increase in sheet resistance. The next comparison, Fig. 5-17, illustrates the effect of retained fluorine dose on the junction depth, measured at $10^{18}/\text{cm}^3$, and sheet resistance of the shallow junction. Clearly, fluorine reduces the transport of boron in silicon while assisting in the activation process. The improvement in both junction depth and sheet resistance is important since it represents a marked change in the activation mechanism. Without an improvement in activation, the R_s - X_j curve (see Fig. 5-23) will not shift to the south-west. Instead, altering the diffusion only causes the data points to run up and down the given R_s - X_j anneal technology curve. For example, less diffusion comes at the expense of higher R_s , while lower R_s comes only with greater diffusion.

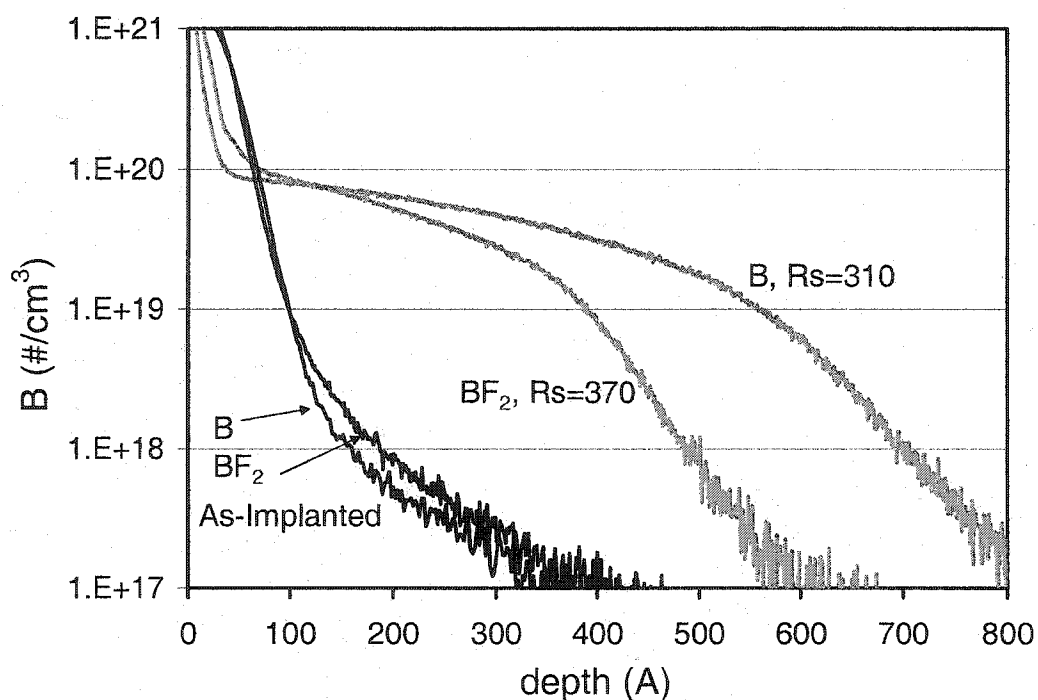


Fig. 5-16. A comparison of 2200 eV, $10^{15}/\text{cm}^2$ BF_2 and 500 eV, $10^{15}/\text{cm}^2$ B diffusion in Si during a microwave spike anneal to 1050°C . The F clearly hinders diffusion.

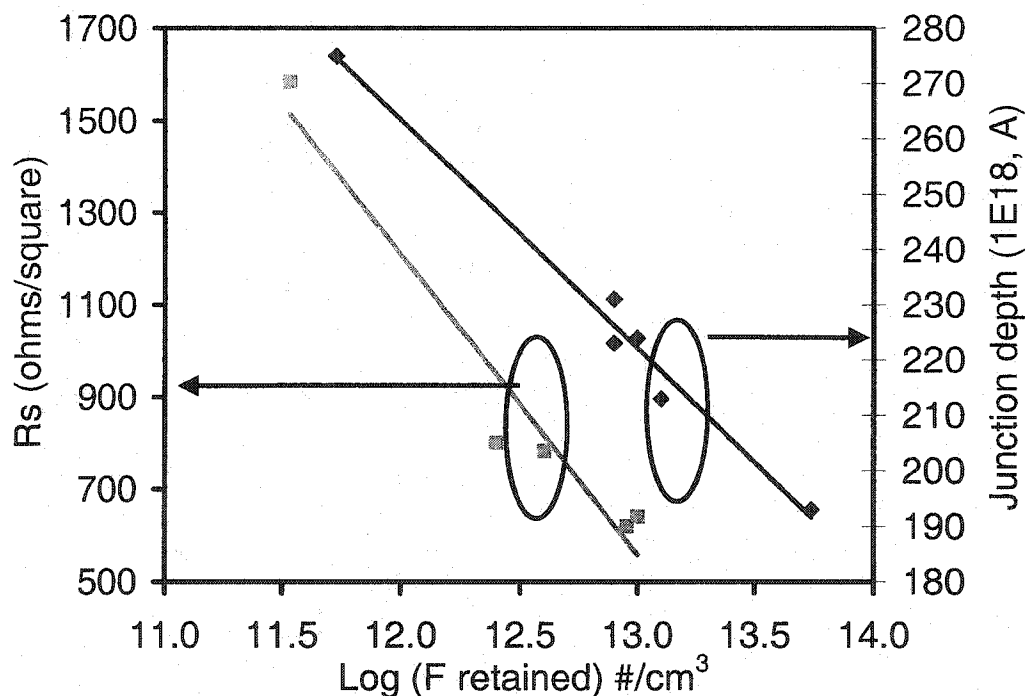


Fig. 5-17. The effect of retained F dose on junction depth ($10^{18}/\text{cm}^3$) and sheet resistance. These data points correspond to Figs. 5-5 and 5-9.

A probable explanation for the lower level of net boron diffusion observed when F is co-implanted with B is presented through the following defect reactions:



In eqn. 5-16, the B_I interacts with the free F to create a B-F associate. These pairs, due to their relatively larger size, have a mobility that is distinctly lower than that of free B_I . The diffusivity (diffusion transport coefficient) of these pairs is markedly lower than the diffusivity (diffusion transport coefficient) of free B_I . This adds another term, D_{B-F} , to eqn. 5-7:

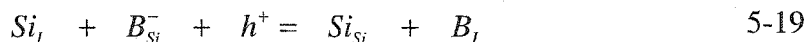
$$J_B(x) = -D_{B,S} \frac{dC_S(x)}{dx} - D_{B,I} \frac{dC_I(x)}{dx} - D_{B-F} \frac{dC_{B-F}(x)}{dx} \quad 5-18$$

Because D_{B-F} is lower than D_I it plays only a small direct role in the transport of boron through silicon. However, it directly lowers the overall $[B_I]$ (B_I transport coefficient), thereby reducing the magnitude of the fast interstitial diffusion, D_I . The result, Figs. 5-16 and 5-17, is a reduction in B diffusion.

Equation 5-17 presents another, more complicated, scenario for reducing boron diffusion. As explained in Sec. 5-3, the creation of a V_{Si} is not possible without the simultaneous creation of a Si_I .



A natural driving force exists for this Si_I to retake its place in the Si lattice by consuming a vacancy. This creates a reaction, as discussed in Sec. 5-3, commonly known as "kick-out", that competes with B_I atoms for silicon sites:



$$B_I = Si_I B_{Si}^2 \exp(-E_{kick}/kT) \quad 5-20a$$

$$B_{Si} = \sqrt{\frac{B_I}{Si_I}} \exp(E_{kick}/2kT) \quad 5-20b$$

The defect reaction of eqn. 5-19 and the mass action laws, eqn. 5-20 a and b, illustrate that Si_I directly reduce the population of active boron atoms. The energy requirements for eqn. 5-12 and 5-19 are approximately 1 eV and -3.9 eV respectively. The exothermic nature of eqn. 5-19 describes a deactivation mechanism that pushes B_{Si} atoms back into interstitial sites. This increase in $[B_I]$ directly increases the magnitude of the fast interstitial component (or the diffusion transport coefficient), eqn. 5-7 or eqn. 5-18, and causes the diffusion process to exhibit a perceived enhancement as discussed in Sec. 5-3. The interaction of F with the Si_I to form the Si-F bonded pair reduces the effective concentration of the Si_I population. This directly reduces the magnitude of the Si_I "kick-out" mechanism, maintains a higher $[B_{Si}]$, and lowers the magnitude of the fast interstitial component of eqn. 5-18. The result is a reduction in the *perceived* enhanced diffusion as Si_I annihilate V_{Si} and deactivate B_{Si} . This reduction in *perceived* enhanced diffusion is witnessed experimentally as a reduction in boron transport for a given thermal treatment.

It is difficult to determine whether the defect reaction described by eqn. 5-16 or eqn. 5-17 dominates during annealing. It is likely that both the B-F and Si-F associates exist and participate in forming a superior junction. It is also likely that the lifetime of these bonded associates is short relative to the time scale of the anneal. The F likely interacts with numerous B_I and Si_I atoms during the duration of the high temperature processing. The extent to which one associate dominates over the other, however, has not been determined since it is extremely difficult to experimentally separate these two reactions.

To examine the impact of F on microwave spike annealing, the retained F dose from the 2200 eV BF_2 implant (Fig. 5-5) and the 800 V BF_3 P²LAD implant (Fig. 5-9) were plotted, Fig. 5-18, as a function of the $[O_2]$ in the ambient. Unlike the calculation for retained B dose,

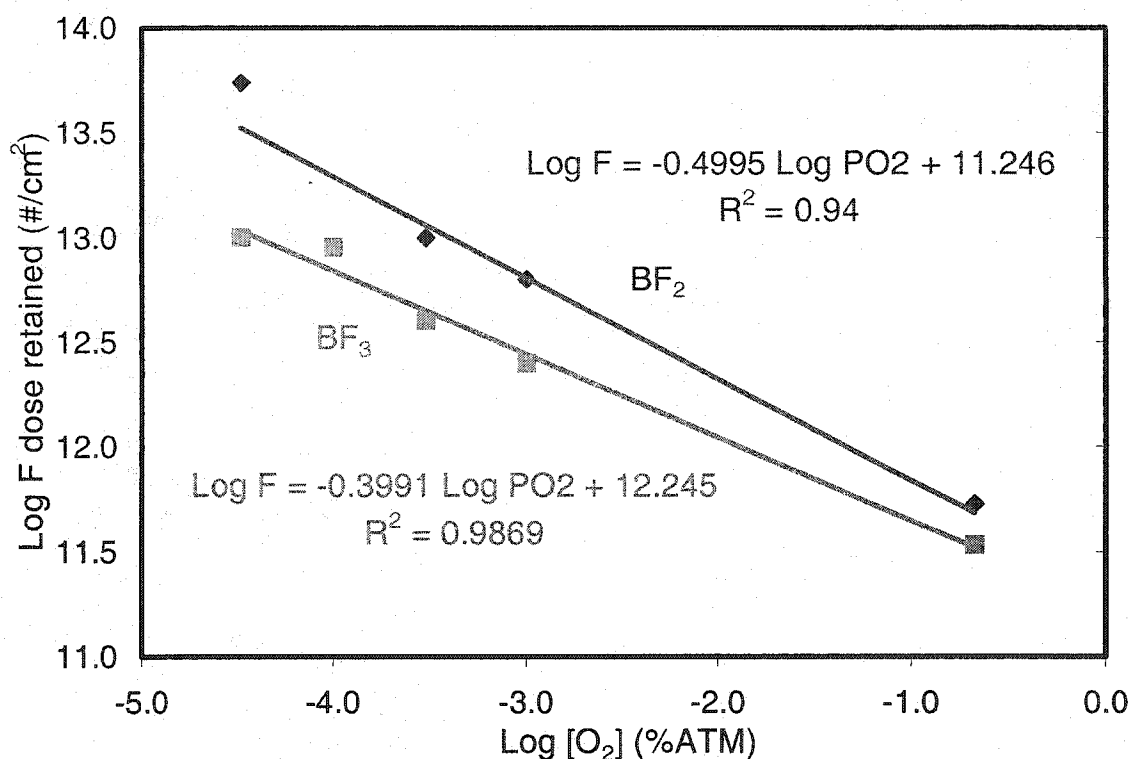


Fig. 5-18. F retained in the silicon for a 2200 eV BF_2 implant, $10^{15}/\text{cm}^2$ and a 800 V P²LAD implant.

which occurred after the oxide strip, the retained F dose was determined by SIMS analysis before the oxide strip, incorporating F both in the surface oxide and in the silicon; thus ensuring that any fluorine loss was due to gasification. The data fit remarkably well to a straight-line plot, with slopes of -0.5 and -0.4 for BF_2 and BF_3 P^2LAD respectively. This indicates a direct link between the $[\text{O}_2]$ in the ambient and the out-diffusion of fluorine. Further proof of the displacement of fluorine by oxygen is shown in Fig. 5-19 and 5-20. These figures show the fluorine and oxygen concentration profiles in silicon as a function of $[\text{O}_2]$ after microwave annealing 800 V BF_3 P^2LAD implants to 990°C . The inward progression of oxygen aligns well with the out-diffusion of fluorine.

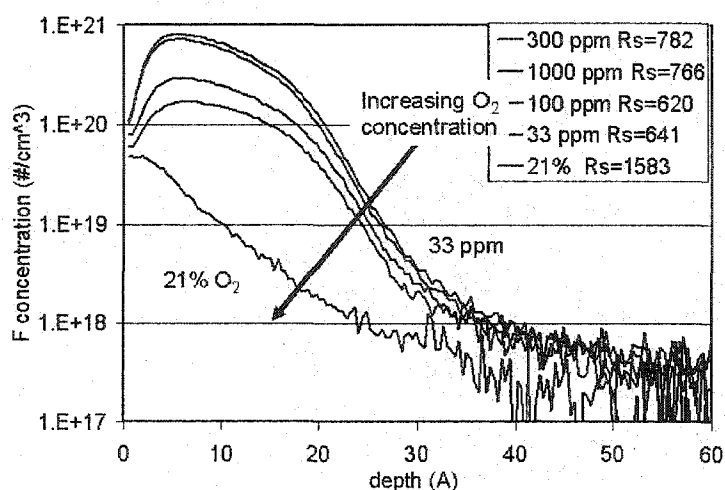


Fig. 5-19. F concentration profiles after microwave annealing an 800 V BF_3 P^2LAD sample to 990°C at various background oxygen concentrations.

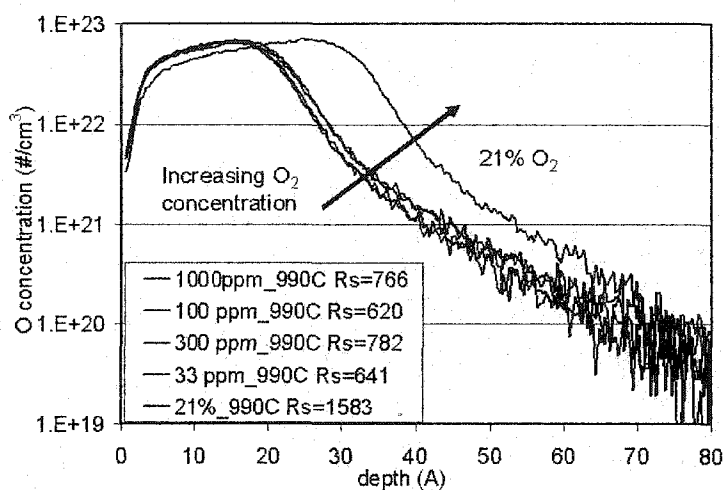


Fig. 5-20. O concentration profiles after microwave annealing an 800 V BF_3 P^2LAD sample to 990°C at various background oxygen concentrations.

While this relationship implies the displacement of F by O_2 incorporation, it does not explain why it occurs. It is, in fact, rather counter-intuitive to propose that a passivating oxide layer on the silicon surface actually enhances the out-diffusion of fluorine from the surface. Two distinct mechanisms, the first chemical and the second electrochemical, are proposed to explain this anomalous behavior. From a chemical perspective, the oxidation of boron rich silicon leads to the formation of both B_2O_3 and SiO_2 . The interaction of F with either B or Si as BF_x or SiF_x is displaced by the oxidation reaction, leaving the F free to out-diffuse. From an electrochemical perspective, the rapid growth of thin oxides on silicon involves the incorporation of trapped positive charge within the oxide layer [40]. Because this trapped charge takes several seconds to decay, a temporary electric field, pointing into the silicon, is created for both p- and n-type implants. Because fluorine is highly electronegative, and because the intrinsic population of electrons in silicon exceeds $10^{18}/cm^3$ at $1000^\circ C$, it is reasonable to speculate that the fluorine atoms will, through electron attachment, negatively ionize and become drawn to the surface by this electric field. Neither one of these models has been adequately tested, so strong statements cannot be made. It is, however, conceivable that either one or both of these reactions plays a strong role in the out-diffusion of fluorine from highly boron doped silicon.

The out-diffusion of F with higher $[O_2]$ in the anneal ambient, Fig. 5-18 – Fig. 5-20, is consistent with the diffusion trends seen in the BF_2 and BF_3 anneals, indicating that the two phenomena are indeed linked. Lamp-based studies have shown that the presence of fluorine has a distinct *chemical* impact on the diffusion and activation of boron [37 – 39]. These microwave studies confirm this chemical impact and illustrate that it is not nearly as straightforward as the lamp-based studies have indicated. The loss of fluorine during microwave

annealing of BF_2 doped silicon at high $[\text{O}_2]$ is consistent with the increased inward diffusion and lower activation of boron shown in Fig. 5-4 – 5-8. The insensitivity of the boron diffusion during microwave annealing of BF_3 P^2LAD implants, Fig. 5-9 and 5-10, and the complete reversal of this diffusion/activation trend during microwave annealing when F is not co-implanted with the boron (i.e. B^+ implants), indicates that fluorine plays a much more complex role in chemically altering the kinetics of diffusion and activation during high temperature processing than was previously presumed.

5.6 Effect of Illumination on Boron Diffusion and Activation

To further illustrate the effect of F in marking the differences between microwave and lamp-based RTP, direct comparison anneals were performed. The results, Fig. 5-21 and 5-22, illustrate the chemical influence of fluorine during microwave annealing. Fig. 5-21 shows the anneal results of a 0.75 keV B^+ implant at a dose of $10^{15}/\text{cm}^2$, and Fig. 5-22 shows the anneal results for a 4 keV BF_2 implant at a dose of $10^{15}/\text{cm}^2$. These samples were microwave and lamp annealed in 100 ppm oxygen to approximately the same temperature, as indicated in the figures. The BF_2 implants show a very similar diffusion profile for both anneal techniques. The B^+ microwave anneals, on the other hand, show a distinct enhanced diffusion mechanism in the extreme near surface region as compared to the lamp-based anneals. The lower sheet resistance of the microwave annealed samples is attributed directly to the more efficient, "box-like" profile (more total dose captured in the junction region) and not to some other enhanced activation mechanism. While the lamp-based anneal results appear less than ideal, it should be noted that the temperature range, $\sim 950^\circ\text{C}$, was chosen as an ideal operating point from the perspective of microwave annealing. A similar "box-like" profile

and R_s values can be obtained with lamp-based RTP, but the target spike anneal temperature is approximately 100°C higher.

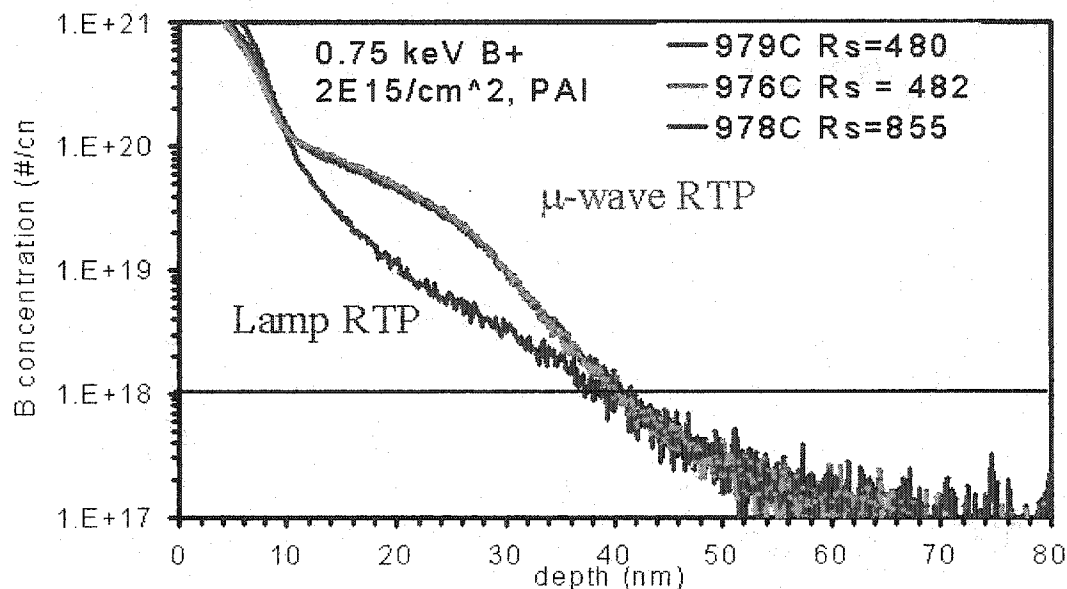


Fig. 5-21. Comparison study of microwave and lamp-based RTP for a 0.75 keV B⁺ beamline implant. A dramatic increase in near-surface diffusion is seen for the microwave case. The [O₂] was ~ 100 ppm.

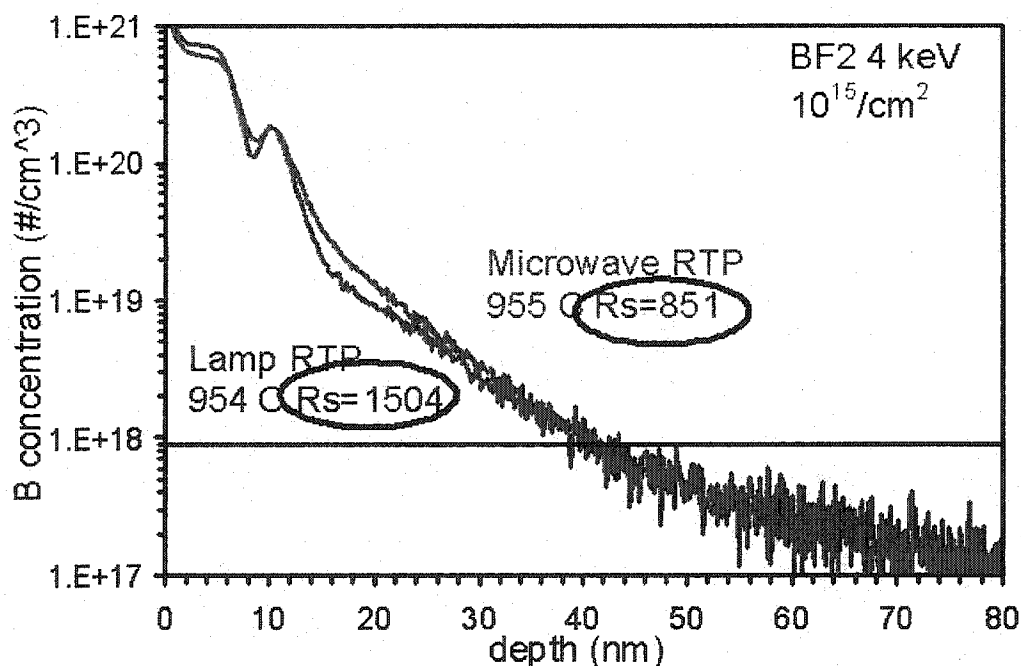


Fig. 5-22. Comparison study of microwave and lamp-based RTP for a 4 keV BF₂ beamline implant. Little difference is seen. The [O₂] was ~ 100 ppm.

It can be concluded from these experiments that the behavior of boron during microwave and lamp-based RTP changes in very different ways when chemically active impurities are either introduced or suppressed. When large amounts of fluorine ($>10^{13}/\text{cm}^2$) are present, the chemical impact of the fluorine appears to dominate the diffusion mechanism regardless of the anneal technique employed. When no fluorine is present, however, microwave annealing in a low, 100 ppm $[\text{O}_2]$ ambient results in increased near surface diffusion. These effects may be caused by the presence of high frequency electric fields in the silicon or by the lack of optically driven photo-generation effects due to the "dark" or non-illuminated nature of microwave annealing.

A comparison study by Fair [42] shows that the diffusion of boron through SiO_2 is enhanced during lamp-based heating as compared to gas furnace heating. He attributes this enhancement to the presence of the high intensity optical illumination present during lamp-based heating. The co-implantation of fluorine resulted in an increase in the diffusion of boron through SiO_2 during gas furnace annealing, although this increase in diffusion was less than the increase seen during the lamp-based diffusion experiments. When the same lamp-based, gas furnace-based heating comparisons were repeated with the fluorine co-implanted samples, the addition of fluorine was found to eliminate the enhancement effects of the high intensity optical illumination [42]. As a result, the penetration of boron through SiO_2 – during lamp-based RTP – in the presence of fluorine is actually *less* than that of boron alone.

If optical illumination affects the transport of boron through SiO_2 , and if fluorine blocks this effect, it is reasonable to conclude that optical illumination will affect the diffusion of boron in silicon and that the presence of fluorine could similarly block this effect. Qualitatively, this appears to be the case in Figs. 5-21 and 5-22. However, there are

unexplained differences as well. From the data in Figs. 5-21 and 5-22, one might hypothesize that optical photons and F inhibit the enhanced diffusion of boron when annealing in a low ppm $[O_2]$ environment. Thus, when F is absent and when the annealing is performed in a low ppm $[O_2]$ environment, there is less enhanced diffusion for lamp-based RTP than for microwave annealing. When F is present, the enhanced diffusion is suppressed equally well for both samples regardless of the $[O_2]$. This suppression may not continue for higher temperature or longer time-scale anneals, since the F is expected to outgas from the wafer during the anneal. The reversal of the oxygen enhanced diffusion trend discussed in Sec. 5.3 is not explained by the lack of high intensity optical illumination, and the differences in diffusion phenomena between microwave and lamp-based systems are not purely driven by an optical mechanism. If this were the case, then the diffusion of B during microwave annealing in the absence of F should be completely insensitive, or at least follow the same trend as BF_2 , to the $[O_2]$. The fact that the trend is reversed, indicates that the high frequency, oscillating electric fields present in the silicon during microwave annealing play an important role in the diffusion of boron that has yet to be determined. This may indicate the existence of a microwave ponderomotive force [43], caused by the presence of high intensity, rapidly oscillating electric fields within the silicon and is a subject of further research.

Another possibility is that the presence of high energy photons affects the evolution of defects during the anneal. It is well known that the long-term degradation of III-V lasers is caused by the evolution of defects over the lifetime of the laser. These dark line defects, DLDs, begin at the cleaved edge of the semiconductor laser and propagate inward during laser operation as high energy photons impact the defect structures [44,45]. Compared to the

thermal energy present, these photons provide a large amount of energy, see cartoon in Fig. 5-23, and stimulate the formation of dislocation loops formed by the collapse of interstitial clusters. As these defects grow, the efficiency of the laser diminishes until reaching failure. It has also been shown [46] that the presence of impurities, for example In impurities in GaAs, can prohibit the growth of these defects. If optical illumination affects the

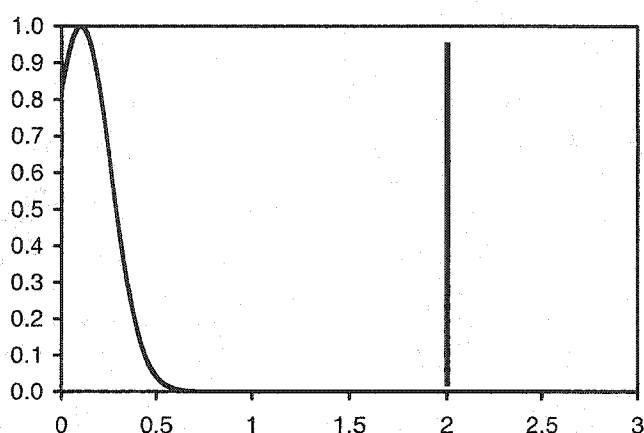


Fig. 5-23. A comparison of the thermal energy present in a silicon wafer when it is heated to 1050°C and the energy provided by photon illumination from a laser at 620 nm. The thermal energy follows a gaussian distribution and falls off exponentially from its peak of 0.9 eV. The photon illumination forms a delta function energy distribution at 2 eV.

evolution of defects in silicon, for example by promoting the dissolution of {311} interstitial clusters, the diffusion of boron in silicon should be noticeably changed. While Fig. 5-21 and 5-22 show noticeable differences in diffusion between microwave and lamp-based RTP, optically enhancing the creation of free Si_i by dissolving the {311} interstitial clusters should result in an *enhanced* diffusion of boron by the "kick-out" mechanism described in Sec. 5-3. This is not consistent with Fig. 5-21 and 5-22, but the presence of a strong microwave ponderomotive force may alter this analysis. The role that optical illumination plays in the evolution of defects is an important area of future experimental research. Future studies, however, must decouple the effect of optical illumination from the microwave ponderomotive force.

5-7 Ultra-Shallow Junction Comparison

Because microwave and lamp-based RTP anneal techniques react differently to secondary chemical impurities, they naturally have different optimization points. For example, it appears that optimal B profiles are achieved with microwave annealing at temperatures that are $\sim 100^{\circ}\text{C}$ lower than with lamp-based anneal techniques. Despite this difference, the overall junction formation appears to be very similar. The $R_s\text{-}X_j$ data curve, Fig. 5-24, shows that the best microwave and lamp-based RTP data points satisfy the 90 nm technology node. The best points for both microwave and lamp-based RTP also occur for the BF_2 and $\text{P}^2\text{LAD BF}_3$ implants. This is yet another indication of the fluorine dominated diffusion mechanism, and future studies involving the co-implantation of F should provide additional insight into this mechanism as well as move the curve further south-west. A comparison of all anneal technologies, Fig. 5-25, indicates that microwave RTP is surpassed only by millisecond lamp-based RTP and laser annealing. Significant improvements in anneal technology are expected when microwave annealing is performed on a time scale of milliseconds.

It is clear from this analysis that those factors which reduce the $[\text{V}_{\text{Si}}]$ and increase the free $[\text{B}_\text{I}]$ atoms enhance the rate of transport of B in Si since the mobility, and therefore the diffusivity, of B_I is far greater than that of B_{Si} . This often times results in the perceived enhancement of boron diffusion down a concentration gradient, for example during the first few seconds of a typical RTP process. The consumption of V_{Si} by oxygen from the ambient and the deactivation of B_{Si} by Si_I are two mechanisms by which this higher level of boron transport, and subsequent perceived enhanced diffusion, is achieved. These mechanisms also negatively affect the electrical carrier concentration since electrical activation is directly

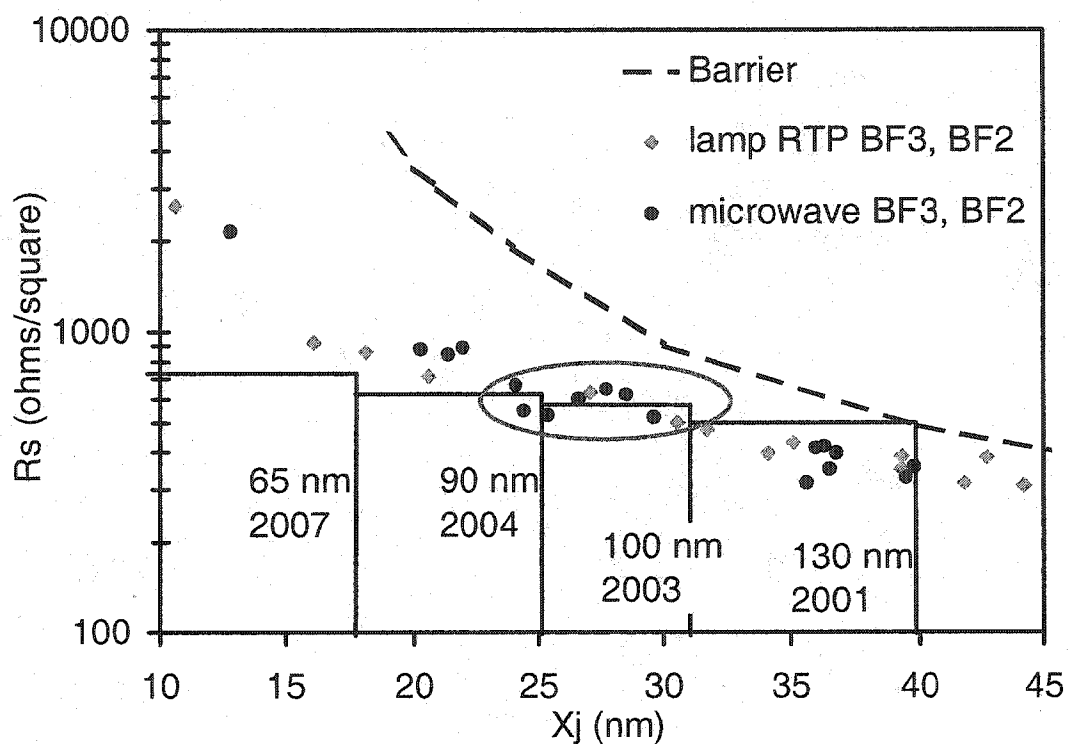


Fig. 5-24. R_s - X_j of microwave and lamp-based RTP.

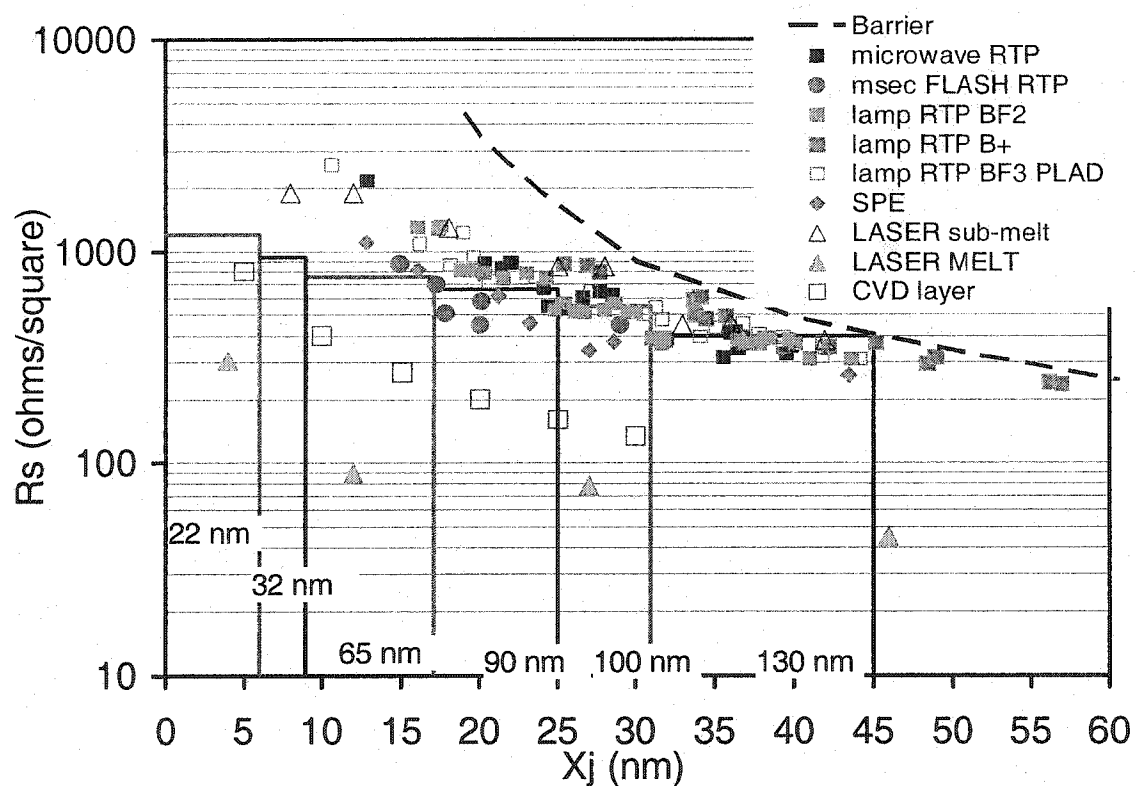


Fig. 5-25. Best R_s - X_j for all anneal technologies.

depend on the $[B_{Si}]$. The presence of certain impurities, such as F, lowers the transport of boron through the creation of defect associates. The creation of B-F complexes directly lowers the overall diffusion by shifting a large fraction of the free B_I population into the less mobile B-F pair. F atoms also tie-up Si_I through a similar Si-F defect associate, thereby reducing the competition for silicon sites. Defect clusters, such as the $\{311\}$ Si_I cluster, play a large role in preventing the deactivation of boron by immobilizing a large number of Si_I atoms. The result is higher boron activation, lower boron mobility, less net diffusion, and a lower sheet resistance. The dissolution of these $\{311\}$ clusters releases Si_I and reverses these positive results as the free Si_I compete with boron for silicon sites. The exploitation of these known phenomena, for example the co-implantation of F during boron implantation and millisecond time scale annealing, will result in superior ultra-shallow junction formation.

CHAPTER 6: BONDING RESULTS

Several <100> silicon wafers were bonded in the microwave and RF systems [45]. Bond uniformity and strength were evaluated by infrared imaging and knife-edge delamination tests. Excellent bonds were obtained in only 5 minutes at 1000°C in both systems. While the time necessary to obtain a full strength silicon-silicon bond is known to be within these time frames [59], conventional furnace processing cannot function on these time scales.

In a typical bonding sequence, the wafers are cleaned first in a NaOH:H₂O₂:H₂O (5:1:1) solution at 80°C, followed by H₂SO₄:H₂O (5:1) at 80°C, and finally dipped in dilute (100:1) HF to remove any residual oxide layer. The wafers are then rinsed in DI water and spun completely dry to minimize the amount of residual moisture on the silicon surface. Once dry, the wafers are immediately pressed together. The initial bond between the two wafers is composed primarily of Si-H-H-Si pairs [60]. Depending on the surface condition of the wafer, (i.e. adsorbed moisture from the atmosphere, surface roughness, etc.), the strength of this bond varies between 200 – 600 mJ/cm². This compares with a theoretical maximum Si-Si bond strength of 2.62 J/cm² [60]. Achieving this high bond strength, requires a high temperature bake (~1000°C) that promotes the out-diffusion of H from the wafer interface and the chemical bonding of the dangling Si atoms.

6.1 Microwave Bonding

To test the effectiveness of microwave heating for silicon wafer bonding, these pre-bonded wafer pairs were ramped at 125°C/sec to 1000°C, soaked for 5 minutes, and passively cooled through radiation losses. Despite the non-uniform nature of microwave heating (see

Ch. 4), an infrared image of the resulting bond, Fig. 6-1, was uniformly void free. The bond strength of the microwave heated wafers was tested using conventional knife-edge delamination tests. At a bonding temperature of 1000°C (wafer center), the wafers shattered before debonding, and the shatter point emanated from the center of the wafer and not at the point of blade insertion. The new surface area created, obtained by summing the perimeter of the shattered pieces (41 cm) minus the original perimeter (23.6 cm) and multiplying by the thickness of the bonded pair (1 mm), was 1.74 cm². When a bonded pair (both wafers of the same material) shatters instead of delaminating, studies by Cook and Gordon [61] state that the minimum bond strength is at least 20% of the strength of the material being bonded. While bond strengths higher than this value are possible, fracture tests can not discern a more precise value. Nevertheless, this minimum bond strength along with the void free IR images indicates a bond of excellent quality.

Assessment of bond uniformity by IR imaging is limited by the available viewing resolution, and defects smaller than this limit go undetected. A more accurate assessment of the largest flaw size involves a calculation of the fracture stress present when the wafer shatters. Multiplying the fracture area (1.74 cm²) and the maximum surface energy of silicon (2.62 J/m²), the dissipated energy is calculated. Because energy is

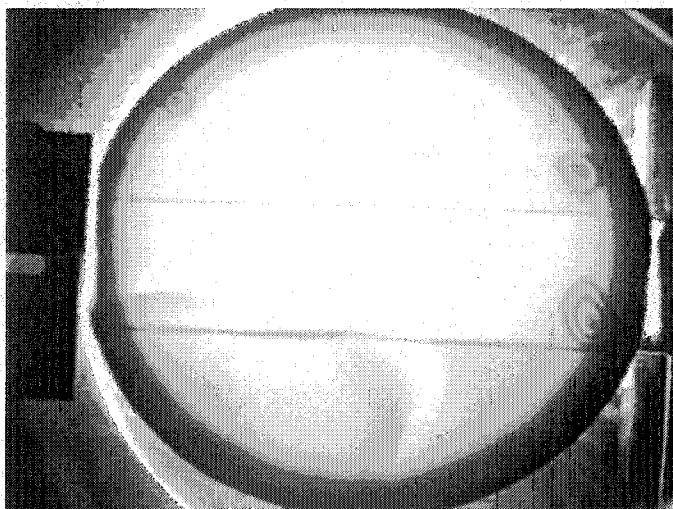


Fig. 6-1. IR images of bonded wafers microwave heated for 5 minutes at 1000°C.

dissipated by mechanisms other than just the shattering process (i.e. kinetic energy), this calculation represents a minimum energy value. Dividing this energy by the volume of the wafer pair gives the minimum energy/unit volume, and this energy/unit volume equates to the minimum fracture stress by the following formula:

$$ENERGY/m^3 = \frac{\frac{1}{2}\sigma_F^2}{E} \quad (6-1)$$

where E is the Young's modulus and σ_F is the fracture stress calculated. The minimum stress for this case is found to be 176 MPa, and with this value the largest defect size is estimated from:

$$TOUGHNESS = 0.9 MPa = \sigma_F \sqrt{c} \quad (6-2)$$

where "c" is the defect radius measured in micrometers. Using this calculation, a maximum defect radius of 26 micrometers is estimated, which is much lower than the resolution available through the IR imaging system used.

The strength of bonds performed in the microwave system at temperatures less than 1000°C was measured to investigate whether microwave annealing offers the possibility of strong bond formation at temperatures lower than those required by conventional heating. Using the formula [60]:

$$\gamma = \frac{3 E t_w^3 t_b^2}{32 L^4} \quad (6-3)$$

it was found that at 900°C the surface energy was ~ 2100 millijoules/m², and at 800°C ~ 700 millijoules/m². Bond strengths below 800°C were not tested. These bond strengths are comparable to those obtained during conventional furnace annealing at the same soak

temperature [59]. These data suggest that for wafer bonding, microwave heating does not provide a lower-temperature advantage over conventional furnace annealing.

6.2 RF Bonding

The significant temperature gradient in the radial direction during RF heating (see Ch. 4) presents an opportunity to study the effect of electromagnetic radiation on materials interactions. During the bonding of the 75 mm wafers, Fig. 6-2, the edge of the wafer was maintained at 1000°C, but the center was only at 780°C. Despite this lower temperature, delamination tests of the silicon bonds at the center of the wafer pair indicate bond strengths typically seen from anneals at temperatures greater than 900°C [60]. The reason these bond strengths were attained at lower temperatures is unclear.

The reaction kinetics involved with silicon bonding may have been enhanced by the high frequency electric fields that are present during heating. Other electromagnetic materials processing experiments have exhibited enhanced reaction kinetics (as compared to conventional thermal techniques) [62,63,64]. Because of the

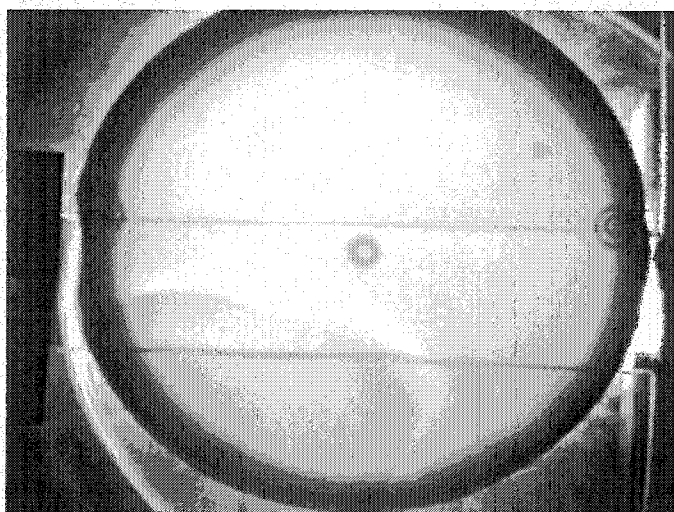


Fig. 6-2. IR images of bonded wafers RF (13.56 MHz) heated for 5 minutes at 1000°C.

shallower skin depth associated with microwave heating (~ 100 microns), these fields would not be present at the bond interface during the microwave bonding experiments. The skin

depth of the radio frequency radiation, however, is approximately 1 mm. Consequently, the electric fields would be present at the bond interface during the RF bonding experiments.

The edge-hot, radial heating pattern may offer an alternative explanation for the strong bond formation at the center of the wafer. Because the edges of the wafer reach 1000°C, the bond will initially set around the edges of the wafer. Once the high-temperature bonding process has begun, the bonding wave-front will propagate inward toward the center of the wafer, which is at a lower temperature than the edges. The surface energy associated with the inward movement of this bond front will provide an additional driving force toward the bonding process. This surface energy increases as the radius of un-bonded area continues to shrink, ensuring that the bond continues to form until the entire wafer is strongly bonded.

6.3 Surface Preparation for Rapid Bonding

Surface preparation is an important part of conventional wafer bonding [59,60]. This is true for EMIH bonding as well, in which poor surface preparation can result in voids, particularly when rapid temperature ramps are used. During the heating process, adsorbed water trapped between the wafer surfaces quickly vaporizes and expands, increasing local pressure along the bond interface. This pressure forces the surfaces apart, creating voids in an otherwise uniform bond. The oxygen released from the water vapor may react locally with the silicon surface to form silicon dioxide in island formations as the bond solidifies across the interface. Once the bond has set, these local oxide growths create voids that are nearly impossible to remove.

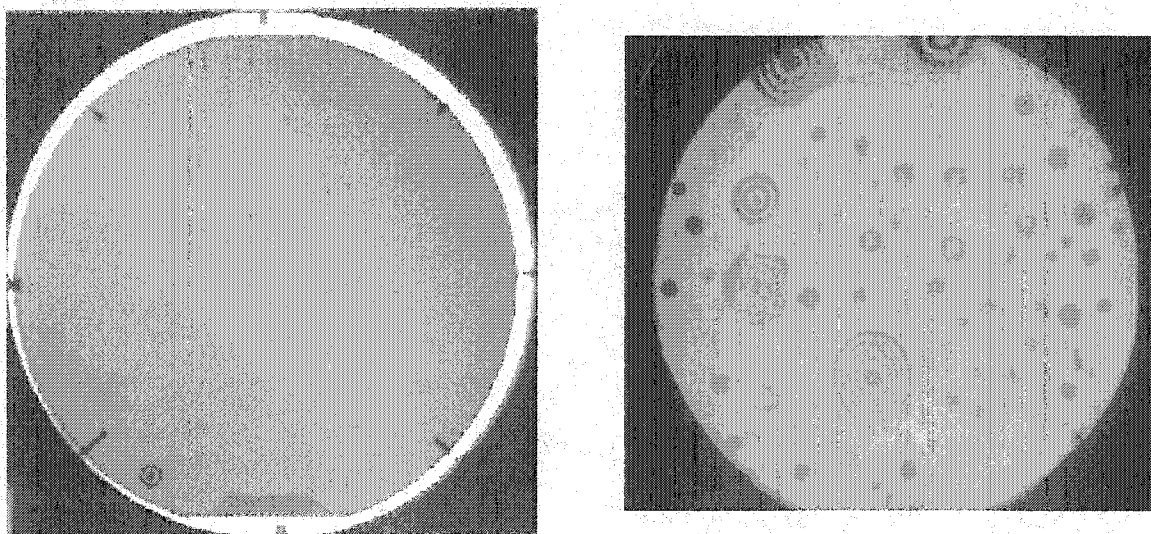


Fig. 6-3. IR images of a wafer pair before (left) and after (right) anneal for 5 minutes at 1000°C. Voids are created from residual water when very high ramp rates are used with either RF or microwave heating.

An increase in the formation of voids (during silicon-silicon bonding) from surface moisture was observed when rapid temperature ramp rates were used to bond a pair of 500 micrometer thick, 100 mm diameter wafers in the microwave system. Prior to bonding, the wafers were cleaned and spun dry, but no HF dips were used to remove residual oxide and no special effort was made to minimize moisture adsorption on the wafer surface. After pressing together, the wafers were ramped to 1000°C at a rate of 125°C/sec, soaked at 1000°C for 5 minutes and passively cooled to room temperature in 5 seconds. Infrared images taken before and after the EMIH cycle clearly show the creation of voids (Fig. 6-3). Voids were not observed when slower ramp rates were used to heat the wafers, and it is presumed that the released gas had enough time to diffuse away from the interface before strongly oxidizing temperatures were reached.

6.4 Silicon-on-Insulator Bonding

As indicated earlier, silicon-on-insulator (SOI) wafer technology may benefit significantly from EMIH bonding, particularly in manufacturing wafers for which the buried oxide layer is relatively thick or located several micrometers below the wafer surface. One method for creating these layers involves growing a thick oxide on a silicon wafer, bonding this wafer to a clean silicon wafer, and polishing back one of the silicon wafers until the desired thickness of silicon remains on the now-buried oxide.

A set of experiments examined the feasibility of SOI bonding with microwave annealing. Several 75 mm diameter wafers with ~ 2500 Å of thermally grown oxide were plasma cleaned, dipped in dilute (100:1) HF for 1 minute, and ultrasonically rinsed in DI water for 5 minutes. The wafers were then pressed together and heated by microwave radiation. Infrared

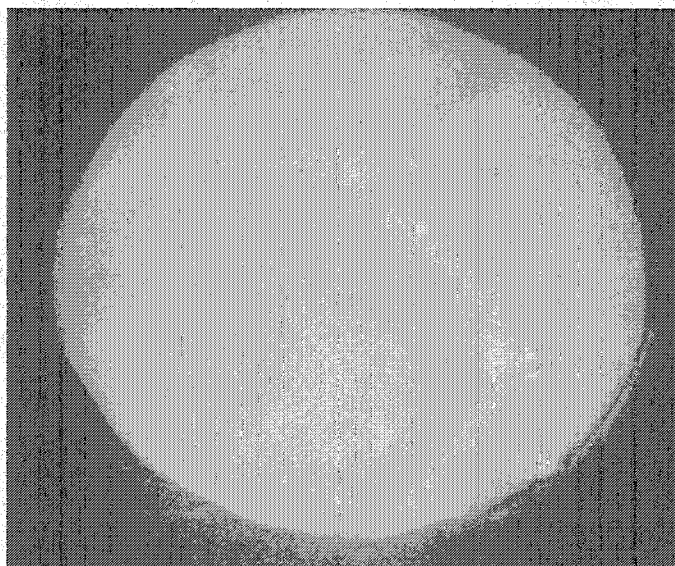


Fig. 6-4. IR images of bonded wafers microwave heated for 5 minutes at 1000°C.

images of the results indicate a very uniform bond (Fig. 6-4). In this experiment the relatively thick layer of oxide prohibits the immediate reaction of oxygen freed from water adsorbed onto the silicon surface, thus eliminating the localized oxidation that occurred in the previous experiments, and allowing the bond to set uniformly.

6.5 Multi-wafer Bonding

In a final set of experiments, multi-wafer bonding was tested in the microwave system. Four pairs of 50 mm diameter wafers, each 250 μm thick, were individually pressed together, stacked on top of each other, and placed in the microwave system. The wafers were heated as in the previous experiments for 5 minutes at 1000°C. Knife-edge tests were unable to delaminate the bonds at either the edges or at any point in the center of the wafers. This experiment demonstrates the potential for multiple stack rapid bonding using EMIH. The use of double-polished wafers should result in a continuous stack of bonded silicon wafers. Unfortunately, the bonding process, which places pressure on the top and bottom outer surfaces of the wafer pair, inevitably damages the top-side of each subsequently bonded wafer. Because the top, outside surface of the first bonded pair is also the bottom inside of the next bonded pair, this damage results in voids during the bonding process.

CHAPTER 7: CONCLUSION

This work has conclusively demonstrated that radiation in the radio (1 MHz – 300 MHz) and microwave (300 MHz – 10 GHz) frequency regimes can efficiently heat silicon. Heating rates of 125°C/sec up to and beyond 1000°C were consistently achieved for 75 mm and 100 mm silicon wafers using 1000 watts and 1300 watts of power respectively.

The theory behind electromagnetic induction heating, EMIH, has been laid out in a straight-forward fashion. The criteria for effective field – wafer coupling and efficient power absorption were described using Maxwell's equations. The effect of materials parameters, such as conductivity, on the heating were explained theoretically, and a functional relationship between radiation frequency and power absorption in silicon was theoretically determined. The experimental apparatuses necessary for achieving the appropriate field configurations for efficient heating were described for both radio and microwave radiation. The heating results for the experimental apparatuses described were reported and the advantages and drawbacks to each design were illustrated both experimentally and through simulations.

The system used to perform heating with radio frequency radiation was found to have a significant radial temperature gradient due to the boundary between the wafer and free space. This gradient rendered this heating system ineffective for the rapid thermal processing of silicon.

Microwave heating was performed in a single mode resonant cavity, and this system proved much more efficient than the RF system. Several resonant modes were found, but heating was performed primarily in the TM₁₁₁ mode. Heating in the microwave cavity was

simulated using Finite Difference Time Domain, FDTD, techniques. These simulations illustrated that the heating was driven by the electric flux transverse to the wafer surface. Unfortunately, the electric flux in a single mode resonant cavity is mode dependent and typically not uniform across the chamber. As a result, the temperature distribution closely matched the non-uniform electric flux distribution. Rotating the wafer during heating averaged out the azimuthal non-uniformity, but a radial non-uniformity persisted.

Full strength silicon bonding requires heating the wafers above 900°C. Because of the rapid and volumetric nature of EMIH, this technique was used to rapidly bond several sets of silicon wafers. The wafers were rapidly heated to 1000°C and soaked for 5 minutes. Excellent, void-free bonding of silicon wafers was achieved in both systems.

The effectiveness of microwave RTP for ultra-shallow junction formation was demonstrated by spike annealing several low-energy boron implanted silicon samples. Sheet resistances of 550 ohms/square with junction depths of ~ 25 nm were achieved using microwave spike annealing. These results demonstrate that microwave RTP is capable of producing junctions very similar to those achieved using lamp-based RTP systems and that it satisfies the 90 nm technology node as dictated by the ITRS technology roadmap. The primary advantage to microwave RTP is that the optimal anneal temperature is approximately 100°C lower than that of lamp-based RTP systems.

The basic kinetics of diffusion and activation during microwave annealing were studied. The background concentration of oxygen in the ambient during the anneal was controlled to gauge the impact of oxygen on junction formation, and fluorine was co-implanted to study its effect on the diffusion and activation of boron in silicon. The results of these experiments were very intriguing. When fluorine was co-implanted with boron, less diffusion occurred

during the anneal and a larger fraction of the implanted boron was activated. The F co-implanted samples also exhibited an oxygen enhanced diffusion effect commonly seen during the lamp-based RTP of boron. As the concentration of oxygen in the system was reduced, the diffusion of boron was reduced and the activation of boron increased. This was attributed to the consumption of silicon vacancies by oxygen atoms that became incorporated in the silicon during the anneal process. When fluorine was not co-implanted with the boron, however, this oxygen enhanced diffusion trend was reversed. As the concentration of oxygen in the ambient during the anneal was decreased, the diffusion of boron in silicon increased. This phenomenon is not observed during lamp-based RTP, and it is attributed either to the lack of high intensity optical illumination during microwave annealing or to some additional microwave ponderomotive force. Comparative anneal studies confirm that a significant difference in the diffusion of boron in silicon exists when the samples are annealed in lamp-based and microwave anneal systems. This difference was not seen when fluorine was co-implanted with the boron.

These intriguing results indicate that the effect of optical illumination on the transport of boron in silicon represents an area of future study. By illuminating a portion of the silicon sample, with a laser whose wavelength falls between 500 – 800 nm, during microwave spike annealing, the effect of high intensity optical illumination on the diffusion and activation of boron in silicon can be directly determined. Controlling the background concentration of oxygen in the ambient during the anneal allows the effect of oxygen and the effect of optical illumination to be decoupled. Performing these experiments on silicon samples that are co-implanted with fluorine presents another avenue of exploration. Finally, the role of optical

illumination on the diffusion of other impurities, such as phosphorous and arsenic, during microwave annealing is a parallel field for future studies.

The presence, and ultimate consequences, of a microwave induced ponderomotive force remains a subject of debate. The lower optimal anneal temperature and the deviation in diffusion trends as a function of oxygen partial pressure when fluorine is not co-implanted in the wafer, with respect to lamp-based systems, provides evidence for a microwave driven ponderomotive force. Future experiments regarding optical illumination should yield more insight into the existence of this ponderomotive force. Because microwave annealing produces junctions that are equivalent, from an R_s - X_j perspective as shown in Fig. 5-23, to lamp-based anneal systems, the presence of this ponderomotive force provides only minor commercial benefits.

While it has been speculated that microwave annealing at the proper frequency might create some sort of resonant interaction with the silicon lattice, and thereby provide some sort of magical activation mechanism, this work clearly illustrates that no such magical resonance exists. In retrospect, this should not come as a surprising result. Even if some novel technique, which provided a discrete amount of energy to every boron atom while leaving the silicon lattice virtually unperturbed, were invented, it is doubtful that it would provide a superior R_s - X_j technology curve. First, phonon interactions within the silicon lattice are so efficient at transferring energy between atoms, time constant on the order of a few hundred femtoseconds [40], that the energy provided to the boron atoms would redistribute uniformly among the silicon lattice before any activation mechanism could occur. This eliminates any advantage in providing the energy only to the boron atoms. Even if this were not the case, the application of energy to an atom does not guarantee that the atom will use the energy,

from the perspective of the engineer providing the energy source, in a productive way. The very same energy needed to drive activation also drives atomic transport through the silicon lattice. The result is both activation and diffusion; running the data points up and down the Rs-Xj technology curve of Fig. 5-23, but not pushing the curve south-west. For such a resonant system to work, the time scale for the entire activation process must be on the order of several hundred femtoseconds. This would prevent the redistribution of energy and restrict the kinetics of diffusion. Unfortunately, it has yet to be shown that several hundred femtoseconds provides enough time for the boron to activate. Furthermore, this time scale is several orders of magnitude shorter than current laser melt/sub-melt technology, raising questions as to whether such a technique is feasible even in a state-of-the-art research lab.

Microwave annealing creates ultra-shallow junctions that are equivalent to those formed during lamp-based RTP. Neither lamp nor microwave annealing technology offers an anneal solution for future technology nodes when the temperature ramp time to 1000°C is on the order of seconds. Microwave annealing on a millisecond time scale, however, does offer a potential solution to the next several technology nodes on the ITRS roadmap. Ramping the wafer temperature to 1300°C in under a millisecond and allowing the wafer to rapidly cool through heat transport to the bulk should eliminate any net diffusion and activate a significant percentage of the implanted dopants. This presents an intriguing area of future study, which will be vigorously pursued over the next few years.

APPENDIX: Derivation of eqn. 2-6

Starting with Maxwell's Equations:

$$\oint E \cdot dl = \iint \frac{d}{dt}(H_\theta) \cdot dS \quad (A-1a)$$

$$\oint E \cdot dl = j\omega\mu \iint H_\theta r \cdot dzd\theta \quad (A-1b)$$

$$\oint H \cdot dl = \iiint I \cdot dv^3 \quad (A-2a)$$

$$\nabla \times H = J_\theta = \frac{\partial H_r}{\partial z} \quad (A-2b)$$

Eqn 1 gives E_θ in the wafer from H_θ

Eqn. 2 gives H_r due to currents induced in the wafer

Apply the geometry of the wafer to get:

$$2\pi a E_\theta = j\omega\mu H_\theta 2\pi a \Delta z \quad (A-1c)$$

$$E_\theta = j\omega\mu H_\theta \Delta z \quad (A-1d)$$

$$H_r = J_\theta \Delta z = \sigma E_\theta \Delta z \quad (A-2c)$$

Setting the total H field, to 1st order as:

$$H_1 = H_o + H_r \quad (A-3a)$$

Inserting H_r :

$$H_1 = H_o + \sigma E_\theta \Delta z \quad (A-3b)$$

Solving eqn. 1d to get the 1st order iteration of E:

$$\vec{E}_\theta^1 = j\omega\mu\Delta z(H^o + \sigma E_\theta\Delta z) \quad (\text{A-4a})$$

$$E_\theta^1 = \frac{j\omega\mu\Delta z}{1 + j\omega\mu\Delta z^2} H^o \quad (\text{A-4b})$$

Allow Δz to equal the wafer thickness, t_w :

$$E_\theta^1 = \frac{j\omega\mu t_w}{1 + j\sigma\omega\mu t_w^2} H^o \quad (\text{A-4c})$$

$$J_\theta^1 = \frac{j\omega\mu t_w}{1 + j\sigma\omega\mu t_w^2} \sigma H^o \quad (\text{A-5})$$

Power density is $J \cdot E$, or σE^2 :

$$P^1 = \frac{\omega^2 \mu^2 t_w^2}{1 + \sigma^2 \omega^2 \mu^2 t_w^4} \sigma H^{o2} \quad (\text{A-6a})$$

Multiply by surface area to get Power:

$$P^1 = \frac{\pi a^2 \omega^2 \mu^2 t_w^2}{1 + \sigma^2 \omega^2 \mu^2 t_w^4} \sigma H^{o2} \quad (\text{A-6b})$$

Let us define a variable δ as:

$$\delta^2 = \frac{2}{\omega \mu \sigma} \quad (\text{A-6b})$$

and re-write power as:

$$E_{\theta}^1 = \frac{j2t_w / \sigma \delta^2}{1 + j(t_w / \delta)^2} H^o \quad (\text{A-4c})$$

$$J_{\theta}^1 = \frac{j2t_w / \delta^2}{1 + j(t_w / \delta)^2} \sigma H^o \quad (\text{A-5})$$

$$P^1 = \frac{4\pi\alpha^2 (t_w / \delta^2)^2}{1 + (t_w / \delta)^4} H^{o2} \quad (\text{A-6b})$$

δ is the classic skin depth, and represents the depth to which the radiation penetrates the material. It should come as no surprise that the maximum power absorption comes when δ equals t_w . When this occurs, no radiation is transmitted and minimum reflection occurs.

LIST OF FIGURES

Fig. 1-1a. A basic MOS device structure. The diode symbols show where the diode effect electrically isolates the doped regions, and the solid arrows indicate where the current and voltage is applied/drawn.

Fig. 1-1b. A MOS device with arrows showing the scaling directions for future devices.

Fig. 1-2. A shallow junction modeled as a resistor

Fig. 1-3. The classis sheet resistance (R_s) junction depth (X_j) tradeoff

Fig. 1-4. Ion implantation process.

Fig. 1-5. Arrhenius plot illustrating the motivation for rapid thermal annealing of the implanted dopants.

Fig. 2-1. Conductivity of silicon as a function of temperature for several different doping levels.

Fig. 2-2. A first order solution for the power absorption efficiency of a silicon wafer obtained through a self-consistent solution of both Ampere's and Faraday's laws.

Fig. 2-3. Overall absorption efficiency of electromagnetic radiation into silicon for the radio, microwave, and millimeter frequency regimes.

Fig. 3-1. Experimental Apparatus: Resonant cavity for microwave heating.

Fig. 3-2. Electric flux patterns, transverse to the wafer surface, for an empty cavity

Fig. 3-3. Relative intensity of the axial directed electric flux in an empty resonant cavity.

Fig. 3-4. Fig. 3-4. a.) support used for steady-state, stationary heating experiments, b.) support used for steady-state, rotational heating experiments, c.) support system used for ultra-rapid annealing of shallow implanted samples.

Fig. 3-5. Experimental Apparatus: Spiral coil design used for RF heating.

Fig. 4-1. Experimental temperature vs. power results at the center and edge of a 7.5 cm silicon wafer heated with the RF spiral antenna illustrated in Fig. 3-5.

Fig. 4-2. Fitted model for the radial temperature non-uniformity which occurs when heating is performed with the RF spiral antenna illustrated in Fig. 3-4.

Fig. 4-3. The heating transient of an intrinsic, 7.5 cm diameter wafer heated at 1000 Watts in the microwave system. The measured heating rate is $125^{\circ}\text{C}/\text{sec}$.

Fig. 4-4. Semilog plot of temperature vs. power for the microwave system.

Fig. 4-5. An illustration of the potential that the microwave cavity has for batch heating.

Fig. 4-6. Pictures of the wafer being heated in the TM111 and TM011 mode while the wafer is rotating at 120 revolutions per minute.

Fig. 4-7. Top down picture of the wafer being heated in the TM111 mode along with quantitative temperature measurements as determined by solid-state reactions.

Fig. 4-8. Uniformity was determined through solid-state diffusion reactions. Wafers were heated with 1000 Watts, Some power loss due to increased reflection during rotation.

Fig. 4-9. Normalized power as a function of time in the simulated cavity after the current source is removed. The decay rate is exponential with a time constant of $5 \times 10^{-9} / \text{sec}$.

Fig 4-10. Simulated axial electric field (E_z) intensity for a 7.5 cm wafer heated in the TM111 mode. The temperature of the wafer is 1050°C , and the field strength has been normalized. The maximum electric field, in the radial range of 3.5 cm from the wafer center, is 4.6 V/cm. The field rises rapidly in the outer 2.5 mm of the wafer, and these values are not accounted for in determining the maximum electric field.

Fig 4-11. Simulated transverse electric field (parallel to the wafer) intensity for a 7.5 cm wafer heated in the TM111 mode. The temperature of the wafer is 1050°C, and the field strength has been normalized. The maximum field strength is 146.8 V/cm, approximately 30 times that of the normal component of the electric field.

Fig 4-12. Simulated temperature map for a 7.5 cm wafer heated in the TM111 mode for 0.6 milliseconds, starting at 0°C. Max Temp = 0.12°C, Min Temp = 0.02°C

Fig 4-13. Simulated steady-state temperature map for a 7.5 cm wafer heated in the TM111 mode. Max Temp = 1050°C, Min Temp = 740°C

Fig 4-14. Simulated electric field intensities for a 7.5 cm diameter wafer heated in the TM011 mode. Because the field (and therefore temperature) pattern is azimuthally symmetric, a single cut through half the wafer is shown. The maximum temperature of the wafer is 1050°C, the maximum transverse electric field is 11.48 V/cm, and the maximum normal electric field is 3.8 V/cm.

Fig 4-15. Simulated electric field intensities for a 7.5 cm diameter wafer heated in the rotational TM111 mode. Because the field, and therefore temperature, pattern is azimuthally symmetric, a single cut through half the wafer is shown. Experimentally measured temperatures are shown as text. The maximum temperature of the wafer is 1050°C, and the maximum electric field is 103.8 V/cm.

Fig. 5-1. An Illustrative concentration profile showing how defects can trap atoms and lead to local concentrations that are much higher than anticipated by eqn. 5-1. In this cartoon, defects at 1600 Å have created a concentration peak.

Fig. 5-2. An example of concentration enhanced diffusion. The fast interstitials diffusers quickly catch up to the slower substitutional atoms. This mobility gradient results in a "box-shaped" profile.

Fig. 5-3. Oxygen enhanced diffusion effects in lamp-based annealing of ^{11}B implants. Anneal step was 10 seconds at 1000°C .

* Figure was reprinted from 32 with permission.

Fig. 5-4. 4500 eV BF_2 implant, $10^{15}/\text{cm}^2$ microwave spike annealed to 1050°C

Fig. 5-5. 2200 eV BF_2 implant, $10^{15}/\text{cm}^2$ microwave spike annealed to 950°C

Fig. 5-6. 2200 eV BF_2 implant, $10^{15}/\text{cm}^2$ microwave spike annealed to 1000°C

Fig. 5-7. 500 eV BF_2 implant, $10^{15}/\text{cm}^2$ microwave spike annealed.

Fig. 5-8. 4500 eV BF_2 implant, $10^{15}/\text{cm}^2$ microwave spike annealed to 1000°C .

Fig. 5-9. 800 V BF_3 implant, $5 \times 10^{15}/\text{cm}^2$ microwave spike annealed to 950°C

Fig. 5-10. 800 V BF_3 implant, $10^{15}/\text{cm}^2$ microwave spike annealed to 990°C

Fig. 5-11. 250 eV and 500 eV $10^{15}/\text{cm}^2$, B implants microwave spike annealed to 950°C

Fig. 5-12. 250 eV and 500 eV $10^{15}/\text{cm}^2$, B implants microwave spike annealed to 1050°C

Fig. 5-13. Log-Log plot of B retained after the microwave spike anneal relative to the oxygen concentration, $[\text{O}_2]$, of the ambient during the anneal.

Fig. 5-14. Atomic interaction by which an oxygen atom consumes a substitutional site.

Fig. 5-15. Oxygen concentration for a 2200 eV BF_2 implant, $10^{15}/\text{cm}^2$ after a microwave spike anneal to 950°C in an ambient whose oxygen concentration was controlled to 100 ppm and in an uncontrolled ambient.

Fig. 5-16. A comparison of 2200 eV, $10^{15}/\text{cm}^2$ BF_2 and 500 eV, $10^{15}/\text{cm}^2$ B diffusion in Si during a microwave spike anneal to 1050°C . The F clearly hinders diffusion.

Fig. 5-17. The effect of retained F dose on junction depth ($10^{18}/\text{cm}^3$) and sheet resistance. These data points correspond to Figs. 5-5 and 5-9.

Fig. 5-18. F retained in the silicon for a 2200 eV BF_2 implant, $10^{15}/\text{cm}^2$ and a 800 V P^2LAD implant.

Fig. 5-19. F concentration profiles after microwave annealing an 800 V BF_3 P^2LAD sample to 990°C at various background oxygen concentrations.

Fig. 5-20. O concentration profiles after microwave annealing an 800 V BF_3 P^2LAD sample to 990°C at various background oxygen concentrations.

Fig. 5-21. Comparison study of microwave and lamp-based RTP for a 0.75 keV B^+ beamline implant. A dramatic increase in near-surface lattice diffusion is seen for the microwave case.

Fig. 5-22. Comparison study of microwave and lamp-based RTP for a 4 keV BF_2 beamline implant. Very little difference is seen for the microwave case.

Fig. 5-23. A comparison of the thermal energy present in a silicon wafer when it is heated to 1050°C and the energy provided by photon illumination from a laser at 620 nm. The thermal energy follows a gaussian distribution and falls off exponentially from its peak of 0.9 eV. The photon illumination forms a delta function energy distribution at 2 eV.

Fig. 5-24. R_s - X_j of microwave and lamp-based RTP.

Fig. 5-25. Best R_s - X_j for all anneal technologies.

Fig. 6-1. IR images of bonded wafers microwave heated for 5 minutes at 1000°C .

Fig. 6-2. IR images of bonded wafers RF (13.56 MHz) heated for 5 minutes at 1000°C .

Fig. 6-3. IR images of a wafer pair before (left) and after (right) anneal for 5 minutes at 1000°C . Voids are created from residual water when very high ramp rates are used with either RF or microwave heating.

Fig. 6-4. IR image of 75 mm wafers with 2500 angstrom of oxide bonded by a 4 minute microwave EMIH anneal at 1000°C.

REFERENCES

1. Y. Chiang, D. Birnie III, W.D. Kingery, Physical Ceramics: Principles for Ceramic Science and Engineering, New York: John Wiley and Sons, 1997.
2. M. Mehrotra, J.C. Hu, A. Jain, S. Hattangady, V. Reddy, S. Aur, M. Rodder, "1.2V, sub-0.09 Micron Gate Length CMOS Technology" in Proc. of International Electron Devices Meeting, 1999, pp. 419-422.
3. P. Packan. "Scaling Transistors into the Deep sub-Micron Regime," MRS Bulletin, pp. 18-20, June 2000.
4. International Technology Roadmap for Semiconductors, found at: <http://public.itrs.net>
5. V. Kasemsuwan, "An Analytical Model of Short Channel MOSFET Including Velocity Overshoot," in Proc. of the 23rd International Conference on Microelectronics, 2002, pp. 451-4, 2002.
6. A.A. Mutlu, N.G. Gunther, M. Rahman, "Analysis of Two-Dimensional Effects on sub-Threshold Current in sub-Micron MOS Transistors," Solid State Electronics, vol.46(8) pp. 1133-7, Aug. 2002.
7. S. Wolf, Silicon Processing for the VLSI era: Volume 4 – Deep submicron Process Technology, Sunset Beach, CA: Lattice Press, May 2002.
8. P.A. Stolk, F.N. Cubaynes, V.M.H. Meyssen, G.Mannino, N.E.B. Cower, J.P. Van Zijl, F. Roozeboom, J.F.C. Verhoeven, J.G.M. Van Berkum, W.M. Van de Wijgert, J. Schmitz, H.P. Tuinhout, P.H. Woerlee, "USJ Formation and Gate Activation in Deep sub-Micron CMOS," in Proc. Material Research Society: Si Front-End Processing – Physics and Technology of Dopant-Defect Interactions II, 2000, vol. 610, pp. B.3.1.1.

9. R. Kasnavi, P.B. Griffin, J.D. Plummer, "A Comparative Study of Dose Loss and Diffusion for B and BF₂ Implants," in Proc. Material Research Society: Si Front-End Processing – Physics and Technology of Dopant-Defect Interactions II, 2000, vol. 610, pp. B.4.3.1.
10. E. Napolitani, A. Carnera, V. Privitera, E. Schroer, G. Mannino, F. Priolo, S. Moffatt, "The Source of TED in sub-keV Implanted Boron in Crystalline Silicon," in Proc. Material Research Society: Si Front-End Processing – Physics and Technology of Dopant-Defect Interactions II, 2000, vol. 610, pp. B.5.2.1.
11. E. Schroer, et al, "TED of Boron at Ultra-Low Energies in Silicon," Appl. Phys. Lett., vol. 76, pp. 3058-60, 22 May 2000.
12. K.S. Jones, H. Banisaukis, S. Earles, C. Lindfors, M. Griglione, M.E. Law, S. Talwar, S.W. Falk, D.F. Downey, A. Agarwal, "Exploring Alternative Annealing Methods for Shallow Junction Formation in Ion Implanted Silicon," in Proc. of the 197th Meeting of the Electrochemical Society: Rapid Thermal and Other Short Time Processing Technologies II, 2000, pp. 119-124.
13. D.F. Downey, S.B. Felch, S.W. Falk, "Doping and Annealing Requirements to Satisfy the 100 nm Technology Node," in Proc. of the 195th Meeting of the Electrochemical Society: Advances in Rapid Thermal Processing, May 1999, pp. 151-162.
14. N.K. Budraa, H.W. Jackson, M. Barmatz, W.T. Pike, J. Mai, "Microwave Induced Direct Bonding of Single Crystal Silicon Wafers," in Proc. 12th IEEE International Conference on Micro Electro Mechanical Systems, 1999, pp.490-2.

15. T.F. Ciszek, "Some applications of Cold Crucible Technology for Silicon Photovoltaic Material Preparation," Journal of the Electrochemical Society, vol. 132, pp. 963-968, 1985.
16. C.A. Balanis, Advanced Engineering Electromagnetics, New York: John Wiley and Sons, 1989.
17. L.C. Shen, J.A. Kong. Applied Electromagnetism, Boston, MA: PWS Publishing, 1995, p.552.
18. K. Thompson, J.H. Booske, R.F. Cooper, Y.B. Gianchandani, "Temperature Measurement in Microwave-Heated Silicon Wafers," in Proc. of Selected Papers from the 2nd World Congress on Microwave Processing, April 2000.
19. K. Thompson, J.H. Booske, Y. Gianchandani, and R. Cooper, "RF and Microwave Annealing for Ultra-Shallow Junction Formation," in Proc. of the 199th Meeting of the Electrochemical Society: Rapid Thermal and other Short-Time Processes II, March 2001, pp. 121-126.
20. R.C. Jaeger, Introduction to Microelectronic Fabrication, Reading MA: Addison-Wesley, 1993.
21. R.F. Pierret, Advanced Semiconductor Fundamentals, Reading MA, Addison Wesley: Reading, MA, 1989.
22. J.D. Jackson, Classical Electrodynamics, New York, NY: John Wiley and Sons, 1999.
23. W. Lerch, B. Bayha, D.F. Downey, E.A. Arevalo, "State of the Art Techniques for Ultra-Shallow Junction Formation," in Proc. of the 199th meeting of the Electrochemical Society: Rapid Thermal and other Short-Time Processing Technologies II, March 2001, pp. 321-326.

24. A. Taflove, S.C. Hagness, Computational Electrodynamics, the finite-difference time-domain method, Boston, MA: Artech House publishing, 2000.
25. E. Jerby, Y. Alpert, "Coupled Thermal-Electromagnetic Model for Microwave Heating of Temperature-Dependent Dielectric Media," IEEE Trans. on Plasma Science, vol. 27, pp. 555-562, April 1999.
26. L. Ma, D.L. Paul, "Experimental Validation of Combined Electromagnetic and Thermal FDTD Model of Microwave Heating Process," IEEE Trans. Microwave Theory and Tech., vol. 43, pp. 2565-2570, 1995.
27. F. Torres, B. Jecko, "Complete FDTD Analysis of Microwave Heating Process in Frequency Dependent and Temperature Dependent Media," IEEE Trans. Microwave Theory and Tech., vol. 45, pp. 108-116, 1997.
28. Y. Bykov, A. Ereemeev, V. Holoptsev, I. Plotnikov, N. Zharova "Spike Annealing of Silicon Wafers Using Millimeter-Wave Power," in Proc. 9th IEEE International Conference on Advanced Thermal Processing of Semiconductors, pp. 82-87, 2001.
29. K. Thompson, J.H. Booske, Y.B. Gianchandani, R.F. Cooper, "Electromagnetic Annealing for the 100 nm Technology Node," Electron Device Lett., pp. 127-9, March 2002.
30. K. Thompson, J. H. Booske, R.F. Cooper, Y.B. Gianchandani, "Electromagnetic Induction Heating for the 70 nm Node," in Proc. Materials Research Society: Silicon Front-End Junction Formation Technologies, April, 2002, vol. 717, pp. C1.3.1 – C1.3.6.
31. K. Thompson, J.H. Booske, R.F. Cooper, Y.B. Gianchandani, " Electromagnetic Fast-Firing for Ultra-Shallow Junction Formation," Trans. Semiconductor Mfg., accepted for publication Feb. 2003.

32. D.F. Downey, J.W. Chow, W. Lerch, J. Niess, S.D. Marcus, "The Effects of Small Concentrations of Oxygen in RTP Annealing of Low Energy Boron, BF_2 and Arsenic Ion Implants," in Proc. Materials Research Society: Silicon Front-End Processing. Physics and Technology of Dopant-Defect Interactions, 1999, vol. 568.
33. T. E. Seidel, D. J. Lischner, C. S. Pai, R. V. Knoell, D. M. Maher, D. C. Jacobson, "A Review of Rapid Thermal Annealing (RTA) of B, BF_2 , and As Ions Implanted into Silicon," Nucl. Instr. and Methods in Physics Research, vol. B7-8, pp. 251-60, March 1985.
34. J.W. Corbett, G.D. Watkins, R.M. Chrenko, R.S. McDonald, "Defects in Irradiated Silicon I. Infrared Absorption of the Si-A Center," Phys. Rev., vol. 121, pp. 1015-1022, 15 Feb. 1961.
35. G.D. Watkins, J.W. Corbett, "Defects in Irradiated Silicon II. Infrared Absorption of the Si-A Center," Phys. Rev., vol. 121, pp. 1001-1014, 15 Feb. 1961.
36. L.S. Robertson, R. Brindos, K.S. Jones, M.E. Law, D.F. Downey, S. Falk, J. Liu, "The Effect of Impurities on Diffusion and Activation of Ion Implanted Boron in Silicon," in Proc. Material Research Society: Si Front-End Processing – Physics and Technology of Dopant-Defect Interactions II, 2000, vol. 610, pp. B5.8.1-6.
37. L.S. Robertson, P.N. Warnes, K.S. Jones, S.K. Earles, M.E. Law, D.F. Downey, S. Falk, J. Liu, "Junction Depth Reduction of Ion Implanted Boron in Silicon Through Fluorine Ion Implantation," in Proc. Material Research Society: Si Front-End Processing – Physics and Technology of Dopant-Defect Interactions II, 2000, vol. 610, pp. B4.2.1-6.
38. D.F. Downey, J.W. Chow, E. Ishica, K.S. Jones, "Effect of F on the Diffusion of Boron in Ion Implanted Si," Appl. Phys. Lett., vol. 73, pp. 1263-5, 31 Aug 1998.

39. D.F. Downey, E.A. Arevalo, K. Thompson, "Fluorine Co-implantation Schemes for 65 nm and Below Ultra-Shallow Junctions," in Proc. of the 203rd meeting of the Electrochemical Society: Advanced Short-Time Thermal Processing for Si Based CMOS Devices, May 2003.
40. Sheng. S. Li, Semiconductor Physical Electronics, New York, NY: Plenum Press, 1993, p.135.
41. R.B. Fair, "Anomalous B Penetration Through Ultra thin Gate Oxide During Rapid Thermal Annealing," IEEE Electron Device Lett., vol. 20(9) pp. 466-468, Sept. 1999.
42. J.H. Booske, R.F. Cooper, S.A. Freeman, K.I. Rybakov, V.E. Semenov, "Microwave Ponderomotive Forces in Solid-State Ionic Plasmas," Physics of Plasmas, vol. 5(5), pp. 1664-70, May 1998.
43. K. Thompson, Y.B. Gianchandani, J.H. Booske, R.F. Cooper, "Direct Si-Si Bonding by Electromagnetic Induction Heating," Journal of Micro-Electro-Mechanical Systems, vol. 11(4), pp. 285-92, Aug. 2002.
44. R.G. Waters, D.P. Bour, S.L. Yellen, N.F. Ruggieri, "Inhibited dark-line defect formation in strained InGaAs/AlGaAs quantum well lasers," IEEE Photonics Technology Lett., vol. 2(8) pp. 531-3, Aug. 1990.
45. R.G. Waters, R.K. Bertaska, "Dark-Line Observations in Failed Quantum Well Lasers," Appl. Phys. Lett., vol. 52(16) pp. 1247-8, 18 April 1988.
46. P.M. Petroff, "Device Degradation and Recombination Enhanced Defect Processes in III-V Semiconductors," Semiconductors and Insulators, vol. 5(3-4) pp. 307-319, 1983.

47. J. Zhu, T.D. Rubia, L.H. Tang, C. Mailhot, G.H. Gilmer, "Ab Initio Pseudopotential Calculations of B diffusion and Pairing in Si," Phys. Rev. B, vol. 54(7), pp. 4741-4747, 15 Aug 1996.
48. H. Schmalzried, Solid State Reactions, pp. 59-61, Deerfield Beach, FL, Verlag Chemie, 1981.
49. D.R. Gaskell, Introduction to the Thermodynamics of Materials, Washington DC, Taylor & Francis Publishing, 1995.
50. D.F. Downey, S.D. Marcus, J.W. Chow, "Optimization of RTA Parameters to Produce Ultra-Shallow, Highly Activated B⁺, BF₂⁺, and As⁺ Ion Implanted Junctions," J. Electronic Materials, vol. 27(12), pp. 1296-1314, 1998.
51. D.F. Downey, C.M. Osburn, J.J. Cummings, S. Daryanani, S.W. Falk, "Dose-rate Effects on the Formation of Ultra-Shallow Junctions with Low-Energy B⁺ and BF₂⁺ Ion Implants," Thin Solid Films, vol. 308, pp. 562-569, 1997.
52. D.F. Downey, D. Brown, J.J. Cummings, A.F. Bertuch, E. Ishida, C. Ng, "Advanced Processing Technique to Minimize Enhanced Diffusion in sub-keV Boron Implants," in Proc. of the 12th International Conference on Ion Implant Technology, June 1998, pp. 55-
53. S. Felch, J. Borland, Z. Fang, J. Koo, "Optimized BF₃ P²LAD Implantation with Si-PAI for Shallow, Abrupt and High Quality P⁺/N Junctions Formed Using Low Temperature SPE Annealing," in Proc. of the 12th International Conference on Ion Implant Technology, Sept. 2002.
54. J.J. Yue, L. Jinning, S. Mehta, K. Jones, "Study of Reverse Annealing Behaviors of p⁺n Ultrashallow Junction Formed Using Solid Phase Epitaxial Annealing," J. Vacuum Science and Technology B, vol. 20(1) pp. 422-6, Jan 2002.

55. D.F. Downey, S.W. Falk, A.F. Bertuch, S.D. Marcus, "Effects of "Fast" Rapid Thermal Anneals on sub-keV Boron and BF_2 Ion Implants," J. Electronic Materials, vol. 28(12), pp. 1340-1344, 1999.
56. J. Gelpey, K. Elliott, D. Camm, S. McCoy, J. Ross, D. F. Downey, E. A. Arevalo, "Advanced Annealing for Sub-130nm Junction Formation," in Proc. of the 201st meeting of the Electrochemical Society: Rapid Thermal and Other Short-Time Processing Technologies III, May 2002, p. 2002-11.
57. J. Ross, S. McCoy, K. Elliott, J. Gelpey, D.F. Downey, E.A. Arevalo, "Characterizing Implant Behavior During Flash RTP by Means of Backside Diagnostics," in Proc. of the 10th International Conference on Advanced Thermal Processing of Semiconductors -- RTP 2002, pp. 99-105.
58. Private communications with Dan Downey, Varian Semiconductor Equipments Associates.
59. M.A. Schmidt, "Silicon Wafer Bonding for Micromechanical Devices," *Solid-State Sensor and Actuator Workshop*, Hilton Head, SC, pp. 127-131, June 1994.
60. Tong, Q-Y and Gosele, U., Semiconductor Wafer Bonding, New York, NY: Wiley Interscience, 1999.
61. J. Cook, J.E. Gordon, in Proc. of the Royal Society, London, 1964, vol. A282 pp.508-520.
62. K.I. Rybakov, V.E. Semenov, "Possibility of plastic deformation of an ionic crystal due to the nonthermal influence of a high-frequency electric field," Phys. Rev. B, vol. 49, pp. 64-68, Jan. 1994.

63. K.I. Rybakov, V.E. Semenov, "Possibility of microwave-controlled surface modification," Materials Research Symposium, vol. 430, pp. 435-440, 1996.
64. K. I. Rybakov, V.E. Semenov, S.A. Freeman, J.H. Booske, R.F. Cooper, "Dynamics of Microwave Induced Currents in Ionic Crystals," Phys. Rev. B, vol. 55(6) pp. 3559-3567, Feb. 1997.