Back-contacted back-junction silicon solar cells

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THESIS

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"The armies separated; and, it is said, Pyrrhus replied to one that gave him joy of his victory that one more such victory would utterly undo him. For he had lost a great part of the forces he brought with him, and almost all his particular friends and principal commanders; there were no others there to make recruits, and he found the confederates in Italy backward. On the other hand, as from a fountain continually flowing out of the city, the Roman camp was quickly and plentifully filled up with fresh men, not at all abating in courage for the loss they sustained, but even from their very anger gaining new force and resolution to go on with the war."

Lucius Mestrius Plutarchus, 46-120 AD
Preface

This document concludes my nearly four years as a Ph.D. student at the Institute for Energy Technology (IFE), Department of Solar Energy. It also summarizes four years of intense learning and much frustration. We set off with the aim of making high-efficiency back-contacted back-junction silicon solar cells. Though a bit disappointing, we soon realized that this was too ambitious considering the time frame, the size of the project, and the available equipment at the time. Nevertheless, I think we ended up with some very interesting results, most of which is presented herein.

Several people have contributed to the scientific content of this thesis. Even more people have contributed to my social well-being as a Ph.D. student at IFE. I would therefore like to begin with thanking all you fine people at the Department of Solar Energy for making my stay at IFE a pleasant one. I would like to thank you for a remarkably large number of coffee breaks of respectful lengths, and for a decent number of "solar beers" at, some times, not so decent locations. Thank y’all.

On the scientific side, I would first and foremost like to thank my main supervisor, Sean Erik Foss. Your competence within an impressive lot of different fields together with your patience, kindness, and helpfulness have been invaluable for the conclusion of this thesis. I would also like to thank my three subsidiary supervisors, Aasmund Sudbø, Tor Arne Fjeldly, and Erik Stensrud Marstein. You have all contributed in different ways at different times in the process, and your help has been greatly acknowledged. Birger Retterstøl Olaisen is the fortunate combination of really clever and really helpful, and for that I am really grateful. I would like to thank Annett Thøgersen for performing transmission electron microscopy (TEM) for me. Jo Gjessing was most dedicated when performing ray tracing, a necessary piece to conclude the device simulation model. Jeyanthinath Mayandi was somehow always present in the lab, and always very helpful. I would also like to thank the head of the department, Arve Holt, for giving me the opportunity of pursuing this Ph.D., which was gratefully financed by the Research Council of Norway.
Last, but most of all I would like to thank my lovely partner, Michelle. I could not have done this without your loving support, patience, and compassion.

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Krister Mangersnes
Abstract

Conventional silicon solar cells have a front-side contacted emitter. Back-contacted back-junction (BC-BJ) silicon solar cells, on the other hand, have both the complete metallization and the active diffused regions of both polarities on the back-side. World-record efficiencies have already been demonstrated for this type of cell design in production, both on cell and module level. However, the production of these cells is both complex and costly, and a further cost reduction in fabrication is needed to make electricity from BC-BJ silicon solar cells cost-competitive with electricity on the grid ("grid-parity").

During the work with this thesis, we have investigated several important issues regarding BC-BJ silicon solar cells. The aim has been to reduce production cost and complexity while at the same time maintaining, or increasing, the already high conversion efficiencies demonstrated elsewhere. This has been pursued through experimental work as well as through numerical simulations and modeling. Six papers are appended to this thesis, two of which are still under review in scientific journals. In addition, two patents have been filed based on the work presented herein.

Experimentally, we have focused on investigating and optimizing single, central processing steps. A laser has been the key processing tool during most of the work. We have used the same laser both to structure the backside of the cell and to make holes in a double-layer of passivating amorphous silicon and silicon oxide, where the holes were opened with the aim of making local contact to the underlying silicon. The processes developed have the possibility of using a relatively cheap and industrially proven laser and obtain results better than most state-of-the-art laser technologies. During the work with the laser, we also developed a thermodynamic model that was able to predict the outcome from laser interaction with amorphous and crystalline silicon.

Alongside the experimental work, we have developed a two-dimensional BC-BJ silicon solar cell device model. The simulations, which are based on the finite
element method, have been performed with the ATLAS device simulator within the Silvaco simulation framework from Silvaco Inc., USA. The device model has been used to optimize the design of a BC-BJ silicon solar cell based on experimental results obtained during the work with this thesis. The model is able to quantitatively predict the performance of cells with different designs, qualities, and dimensions through optical and electrical simulations, and thereby giving us a good indication of the efficiency potential of the cell structure. It has also given us valuable insight into the physics determining the performance of a BC-BJ silicon solar cell. From this insight, important conclusions regarding the design rules of this type of solar cell devices could be drawn. Finally, the device model was used to investigate quantum mechanical tunneling mechanisms in the junction between the adjacent, highly-doped regions of opposite polarity on the backside of the cell. Through the simulations we found some simple design rules that need to be followed in order to avoid shunting-like behavior due to unwanted trap-assisted tunneling in the lateral tunneling junction. At the same time, band-to-band tunneling entails potential current breakdowns at low to moderate reverse biases. This implies that local hot-spots can be avoided since the heat distribution under reverse bias will be distributed throughout the whole junction area. Thus, by careful optimization and tailoring of the doping profiles, the tunneling may enable the use of back-junction silicon solar cells in a solar module without the need for bypass diodes.
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Chapter 1

Introduction

In 2008, average worldwide power consumption was 15 TW\(^1\). Approximately 86 percent of this was derived from the combustion of fossil fuels[1]. Figure 1.1 shows the development of worldwide energy consumption from 1965 to 2008. The increase in energy consumption is simply a result of the world’s growing population as well as an increase in the average gross domestic product (GDP) per capita[2]. The population is expected to reach 9 billion between 2040 and 2050[3]. In addition, the GDP per capita of a country is known to be directly proportional to its energy consumption per capita[2]. Thus, when taking into account that the average GDP per capita is growing even faster than the population, a drastic increase in the world’s energy demand and consumption seems inevitable in the near future. This prospected increase in energy consumption is one of the biggest challenges faced by our generation. The global petroleum resources are limited. Some claim that the production of oil has already entered a terminal decline, and that proactive mitigation of an energy crisis might not be possible[4, 5]. More optimistic sources claim that peak oil will happen in 2020 or even later, and that it is possible to maintain a high production of oil several decades from now[6]. Nevertheless, the fossil resources are getting less accessible. Oil is pumped out from deeper wells. Oil rigs are being built in distant areas with hostile climate, and more technologically challenging solutions need to be implemented to withdraw the remaining resources. Not only does this increase the price of production, but

\(^1\)This comprises commercially traded fuels only. Excluded, therefore, are fuels such as wood, peat, and animal waste which, though important in many countries, are unreliably documented in terms of consumption statistics. Also excluded are wind, geothermal, and solar power generation.
CHAPTER 1. INTRODUCTION

Figure 1.1: Worldwide energy consumption from from 1965 to 2008. The units are in million tons of oil equivalent. (Source: BP statistical review of world energy 2009[1]).

also the risk of serious oil spills like the Deepwater Horizon oil spill in the Gulf of Mexico in 2010[7]. Action needs to be taken to ensure a sustainable energy supply in the future, and there is already a strong ongoing political effort to allocate energy consumption away from fossil fuels. In addition, there has been much focus on the role of fossil fuels on global warming. According to the Intergovernmental Panel on Climate Change (IPCC) most of the observed temperature increase since the middle of the 20th century was very likely caused by increasing concentrations of greenhouse gases in the atmosphere resulting from human activity, such as fossil fuel burning and deforestation[8]. This has greatly increased focus on renewable energy both from a scientific and political point of view, and several sources of renewable energy are experiencing a significant growth in investments and installed capacity[9]. At the beginning of 2009, the total production capacity of renewable energy was 1.1 TW$^2$, with hydroelectric as the dominating source with a share of 83 %[9]. When evaluating the potential of different sources of renewable energy:

$^2$This includes all power generating sources of renewable energy, but excludes heating sources such as solar collectors for hot water, geothermal, and biomass. Neither are renewable transport fuels, such as ethanol and biodiesel, included.
renewable energy, one needs to take into account the available resources. Solar energy is, by far, the most abundant energy source. In Figure 1.2 the available total energy from different renewable energy sources are represented as the volume of different cubes and compared to the global average consumption. The available solar resources are about 6000 times the total global energy consumption. Thus, less than 0.02 % of the solar resources are sufficient to entirely replace fossil fuels and nuclear power as an energy source. Figure 1.3 shows worldwide insolation and the possible locations of large solar power plants that in total could provide the total global energy demand. At the end of 2009, the total installed capacity of solar power was 13 GWp\(^3\)[9]. This only accounts for approximately 0.04 % of the total global energy consumption. Still, grid-connected photovoltaic (PV) electricity has been the fastest growing energy source over the last ten years, with an average annual increase of about 42 %[9]. In other words, a doubled installed capacity every second year, see Figure 1.4. To be able to continue this growth, and to make solar power a major supplier in the future energy market, the technology needs to be cost competitive compared to alternative sources of energy.

\(^3\)Wp, or watt-peak, is a measure of the nominal power of a solar cell device under standard illumination and operating conditions; a light intensity of 1000 W/m\(^2\), with a spectrum similar to sunlight hitting the earth’s surface at latitude 35°N in the summer (airmass 1.5) and a temperature of 25 °C.
20 years ago, solar cells were still regarded as a niche market, and interesting only in off-grid and space applications areas where other sources of power either were unavailable or too expensive. Today, the situation is completely changed. Increased cell performance at reduced production and installation costs have cut down energy pay-back times for some solar modules to close to one year[10, 11]. With a further reduction in solar electricity prices, the technology will become cost-competitive in yet larger markets. In a recent roadmap on solar power from the International Energy Agency (IEA), PV is projected to provide 5 % of global electricity consumption in 2030, rising to 11% in 2050, corresponding to an installed capacity of more than 3 TWp[12]. Today, 80 % of the installed capacity of solar electric comes from crystalline silicon solar cells[13]. Despite being a poor optical absorber, silicon is by far the most commonly used base material for solar cells. The combination of abundance, availability, non-toxicity, and conve-

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**Figure 1.3:** Map of average global insolation. The black dots represent possible locations of large solar power plants that in total could provide the world’s total energy demand. (Source:http://www.ez2c.de/ml/solar_land_area/)

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4Though this corresponds to a 250 times increased capacity compared to 2009 values, most of the solar industry and several environmental organizations find this estimate far too modest.
Convenient physical properties make silicon a natural choice of material for PV also in the future[14]. The relatively young PV industry also has benefitted greatly from the huge knowledge base of the microelectronic industry. Still, a necessary road ahead also for silicon solar cells is a further reduction in the price of solar electricity. This can be done either by cutting production and installation costs without hampering cell performance, or by increasing power output without increasing production and installation costs. Preferably both.

One very promising high-efficiency solar cell concept is the back-contact back-junction (BC-BJ) silicon solar cell[15, 16]. This type of cell design has already been proven to be compatible both with high efficiency and large-scale production[17]. In fact, it is the silicon solar cell concept that have demonstrated the highest efficiencies both on cell and module level in production[18, 19]. Still, several issues need to be solved for BC-BJ silicon solar cells, both to increase the efficiency further towards the theoretical limit[20], and to reduce the production costs. In this thesis several important aspects related to BC-BJ silicon solar cells are addressed with the aim of both increasing efficiency and reducing costs. This is done through simulations, modeling, and laboratory experiments.

**Figure 1.4:** Total installed capacity of solar electricity from 1996 to 2008.(Source: BP statistical review of world energy 2009[1])

![Graph showing total installed capacity of solar electricity from 1996 to 2008.](image)
CHAPTER 1. INTRODUCTION

1.1 Thesis outline

Chapter 2 starts with a short introduction to silicon solar cells in general. Subsequently, back-contacted back-junction (BC-BJ) silicon solar cells are presented more in detail. Other types of back-contacted silicon solar cell designs are also briefly mentioned, and a short historic overview is given at the end of the chapter. Chapter 3 comprises the background and theory of the experiments performed during the work with this thesis. The most central equipment used for processing and characterization are presented. Plasma-enhanced chemical vapor deposition (PECVD) and laser technology have been especially central in the experimental work, and will be discussed somewhat more thoroughly. Also included in the chapter is a discussion of laser interaction with crystalline and amorphous silicon. In chapter 4 some experimental results that have not been included in any of the papers are presented. The main experimental results are given in papers II, III, and V, but some work of interest has not been published, and is therefore discussed herein. In general, experimental focus has been set on processes related to the backside of a BC-BJ silicon solar cell, and as will be explained in the next chapter, the front- and backside of a BC-BJ silicon solar cell can be optimized more or less independently. Complete solar cells have not been made during work with this thesis. On the other hand, it has been possible to investigate the influence from single processing steps on the performance of a finished solar cell through device simulations. In chapter 5 the background of the simulations performed with the ATLAS device simulator within the Silvaco simulation framework is presented. A conclusion is given in chapter 6.

Most of the work performed and described in this thesis and the appended papers was done by the author, but with some exceptions. Transmission electron microscopy (TEM) was performed by Annett Thøgersen, Institute for Energy Technology, Department of Solar Energy (IFE-Solar). Secondary ion mass spectroscopy (SIMS) was done by Alexander Azarov, Center for Materials Research, University of Oslo. Capacitance-voltage ($C-V$) measurements were performed by Edouard Monakov, IFE-Solar. Most of the simulation work was also done
1.1. THESIS OUTLINE

by the author, but again with some exceptions. Ray tracing was performed by Jo Gjessing, IFE-Solar, while SPICE module simulations and ATHENA process simulations were done by Sean Erik Foss, also IFE-Solar.

Six papers are appended to this thesis, two of which are still under review in scientific journals. They all cover different and important issues regarding processing and simulations of BC-BJ silicon solar cells.

Paper I gives us valuable insight into the physics determining the performance of a BC-BJ silicon solar cell. From this insight important conclusions regarding the design rules of this type of solar cell devices can be drawn. The paper serves as a starting point for the simulation model, which has been gradually developed alongside with the experimental work during the whole Ph.D. process. Paper II deals with laser structuring of the backside of a BC-BJ silicon solar cell. Optimized laser and etching parameters are demonstrated to enable ablation (removal) of large areas of silicon from a silicon wafer without influencing the material quality in the processed regions, making it possible to produce self-aligned interdigitated finger patterns with flexible designs and dimensions.

In paper III the same laser is utilized to ablate holes in a double-layer of passivating amorphous silicon (a-Si) and silicon oxide (SiO$_2$) without damaging the underlying crystalline silicon (c-Si). This is done with the aim of making local contacts on the back of a BC-BJ silicon solar cell.

To further investigate the interaction between the laser and a-Si and c-Si, a thermodynamic model is developed that is able to predict the laser threshold fluence (energy density) for SiO$_2$ ablation as a function of the thickness of the a-Si layer. The model is presented in paper IV.

In paper V the quality of aluminum contacts deposited through the laser ablated holes discussed in paper III is investigated. The quality of the contacts is first optimized experimentally, before the results are used to numerically optimize the design of a BC-BJ silicon solar cell by device simulations, and so to get an indication of the efficiency potential of the cell design we are working with.
CHAPTER 1. INTRODUCTION

The last paper, VI, is dedicated to numerical investigation of potential quantum-mechanical tunneling of charge carriers in the junction between the adjacent, highly doped regions of opposite polarity on the back of a BC-BJ silicon solar cell. This is done through both device, process, and circuit simulations.
1.2 List of publications


1.3 Filed patent applications


2.1 The basics of silicon solar cells

A solar cell is a device that converts sunlight directly into electricity through the photovoltaic effect\cite{21}. The efficiency of the cell quantifies the fraction of solar energy that it converts into electrical energy. In principle it depends on two parameters. The generation of current by absorbed incident illumination and the loss of charge carriers via so-called recombination mechanisms. A built in p-n diode separates charge carriers of opposite polarity, and drives the light-generated current through the cell and to the terminals. Finally, several solar cells will be electrically connected and encapsulated as a module. In mathematical terms the efficiency, $\eta$, of a solar cell is given as:

$$\eta = \frac{P_{\text{out}}}{P_{\text{in}}} = \frac{FF \cdot J_{SC} \cdot V_{OC}}{P_{\text{sun}}},$$

where $P_{\text{in}}$ is the incoming power density and $P_{\text{out}}$ is the maximum output power density of the cell. $J_{SC}$ and $V_{OC}$ are the short-circuit current density and the open-circuit voltage, respectively, while $FF$ is the solar cell fill factor. $FF$ describes the squareness of the current-voltage ($J$-$V$) curve. $P_{\text{sun}}$ is the power density incident on the cell from the sun. Solar cell $J$-$V$ characteristics are usually measured
under a set of standard test conditions where $P_{\text{sun}}$ is 0.1 W/cm$^2$ of a normally incident air-mass (AM) 1.5 spectrum at 25 °C. A typical $J$-$V$ curve of a high-efficiency silicon solar cell is shown in Figure 2.1. The $J$-$V$-curve of a solar cell is commonly shown in the first quadrant with the light-induced current density defined as positive and the dark current negative. The squareness of the curve is limited by the current’s exponential dependence on voltage in a p-n diode. A one-diode equivalent circuit of a solar cell that also includes parasitic series and shunt resistances is shown in Figure 2.2. The corresponding $J$-$V$ equation is given by

\[ J(V) = J_{SC} - J_0 \exp \left[ \frac{(V + JR_S)}{V_{th}} \right] - \frac{V + JR_S}{R_{SH}}, \]

(2.2)

where $J_0$ is the dark saturation current density, $V$ is the applied voltage, $V_{th}$ is the thermal voltage$^1$, $R_S$ is the series resistance, and $R_{SH}$ is the shunt resistance. $R_S$ and $R_{SH}$ should be as small and large as possible, respectively, in order to maximize the efficiency of a solar cell.

Even if silicon is a natural choice as base material for solar cells, there are theoretical limitations to the possible maximum efficiency. In a single p-n junction

$^1$The thermal voltage is a proportionality factor that relates the electrostatic potential to the electrical current through a p-n diode. It is given as $V_{th} = \frac{kT}{q}$, where $T$ is the absolute temperature and $k$ and $q$ are the Boltzmann constant and the magnitude of the electron charge, respectively.
2.2. BACK-CONTACTED BACK-JUNCTION SILICON SOLAR CELLS

Figure 2.2: A one-diode equivalent circuit of a silicon solar cell, including parasitic series ($R_S$) and shunt ($R_{SH}$) resistances. $J_{SC}$ and $J_0$ are the light-induced and dark current density, respectively.

device, low-energetic photons will pass through the cell without being absorbed. Higher-energetic photons will lose their excess energy, compared to the bandgap, as lattice heat. Energy is also lost across the p-n junction, at the contacts, and due to recombination. These loss mechanisms limit the maximum efficiency of a single-junction silicon solar cell to approximately 30 %, according to Shockley and Queisser[20]. The present world record efficiency is 25 % for a monocrystalline silicon solar cell[22, 23] and 20.4 %$^2$ for a multicrystalline cell[23, 25]. In production the efficiency of state of the art large area modules is typically 3-5 % lower than that of the record cells[17, 23, 26].

2.2 Back-contacted back-junction silicon solar cells

Back-contacted silicon solar cells comprise solar cells where the electrical busbars of both polarities are located on the backside$^3$. In specific, back-contacted

$^2$The cited articles from Zhao et al. and Schultz et al., report efficiencies of 24.7 % and 20.3 % for the mono- and multicrystalline silicon solar cells, respectively. More recently, the solar reference spectrum has been revised[24]. This entailed a 0.3 % absolute increase in efficiency for the monocrystalline cell and 0.1 % for the multicrystalline cell, leading to new world record efficiencies for silicon solar cells of 25 % and 20.4 %, respectively.

$^3$An excellent review of back-contacted silicon solar cells was given by Kerschaver and Beaucarne[15].
CHAPTER 2. BACK-CONTACTED BACK-JUNCTION SILICON SOLAR CELLS

back-junction (BC-BJ) silicon solar cells, which is the main topic of this thesis, have both the complete metallization and the complete contacted diffused regions located on the backside of the cell. It was originally designed with the emitter and back-surface field and the respective contacts forming an interdigitated finger pattern[16]. This interdigitated back-contact (IBC) structuring is still the most common way of aligning the structures on the backside of the cell, and the phrases IBC and BC-BJ are both commonly used to describe this cell design. Herein we will use the latter. A schematic drawing of a BC-BJ silicon solar cell is shown in Figure 2.3. This cell design entails numerous advantages compared to conventional front-contacted silicon solar cells. Moving all contacts to the backside eliminates front-surface contact reflection. Thus, the conventional trade off between reflection and series resistance in the contact fingers is completely omitted. As a consequence, the sunward and rear side can be more or less independently optimized for optical and electrical performance, respectively. The low blue response commonly seen in front-contacted silicon solar cells, due to the high recombination in the front side emitter, is significantly improved. This justifies optical optimization for a broader set of wavelengths, and double-layer anti-reflection coatings are commonly used in combination with conventional alkaline surface

Figure 2.3: Schematic drawing of a back-contact back-junction silicon solar cell. The cell is shown upside-down, and the dimensions are not to scale. FSF and BSF, are the front- and back-surface fields, respectively.
texturing for reduced front-side reflection. BC-BJ silicon solar cells therefore usually have a darker color appearance than conventional silicon solar cells. In combination with the clean, non-metallized front surface, this makes BC-BJ silicon solar cells especially attractive from an architectonic point of view, which is becoming more important with building-integrated photovoltaics. A BC-BJ solar module from Sunpower Corporation is shown in Figure 2.4. Moreover, having

Figure 2.4: A solar module from Sunpower Inc. The module is made from BC-BJ silicon solar cells, and holds the world-record efficiency for a commercially available silicon solar module. (Source:http://www.sunconnex.com/default.asp?page=SunPower-SPR)
contacts of both polarities on the same side may simplify cell interconnection in modules, and allows for an increased packing density of cells[17, 27, 28].

On the processing side, however, there are some challenges. This is mostly related to the closely spaced interdigitated fingers of opposite polarity on the backside of the cell, see Figure 2.3. The processing therefore requires careful alignment of both diffused regions and contacts. Since photolithography is regarded unsuited for cost-effective large-scale production of silicon solar cells, low-cost alternatives, such as screen-printing, are commonly used for one or several alignment steps in BC-BJ research and production[17, 29]. This limits the minimum pitch to the order of mm. This is in contrast to the cells produced in the early stages of BC-BJ silicon solar cell research and production. With the use of photolithography a pitch in the order of 50 µm was easily obtained. This might still be possible to justify for concentrator applications, but for one-sun flat solar panels, photolithography is too expensive as a patterning tool to make the cells cost competitive. Another process-related challenge is the silicon material quality. Since the emitter is located on the backside of the cell, and the charge carriers are generated predominantly near the front surface, the carriers have to diffuse through the whole thickness of the wafer. With a pitch in the order of mm the average required diffusion distance of minority carriers is even more lateral than vertical. The required maximum diffusion length, $L_{\text{max}}$, therefore depends on the wafer thickness, $W$, and the width of the back-surface field, $d_{\text{BSF}}$, see Figure 2.5. $L_{\text{max}}$ can be calculated as

$$L_{\text{max}} = \sqrt{\frac{1}{4} d_{\text{BSF}}^2 + W^2}. \quad (2.3)$$

With a wafer thickness of 200 µm and a BSF width of 600 µm this results in a maximum diffusion length of approximately 360 µm. This is indeed a two-dimensional problem, and device simulations should at least be in two dimensions. The need of long diffusion lengths call for high-quality monocrystalline wafers with long minority carrier lifetimes and excellent surface passivation. The minority carrier lifetime also needs to remain high through the whole process and under illumination. This can be a challenge, but the use of n-type (instead of p-type) wafers has been shown to facilitate the challenges related to process and
2.2. BACK-CONTACTED BACK-JUNCTION SILICON SOLAR CELLS

Figure 2.5: Schematic, simplified, side-view drawing of the smallest repeating unit of a back-contact back-junction silicon solar cell. The diagonal line shows the maximum required diffusion length for minority charge carriers. BSF is the back-surface field. The drawing is not to scale.

light-induced degradation of the silicon material, see section 2.4.

2.2.1 The emitter-wrap-through cell design

The emitter-wrap-through (EWT) solar cell concept is quite similar to the BC-BJ solar cell design[30]. The complete metallization is still located on the back of the cell, but a part of the collecting emitter is now close to the front surface. This lowers the requirements for high minority carrier lifetime materials compared to BC-BJ cells, as the average diffusion length is significantly reduced. The extension to the backside contacted junction is done through laser drilled vias in the wafer. On a 5 inch square wafer, between 10000 and 20000 holes are typically used as transporting channels for minority carriers collected at the front or through the complete thickness of the silicon wafer. The EWT cell concept includes almost all the advantages of a BC-BJ silicon solar cell, but with reduced requirements for material quality. A scanning-electron microscope (SEM) picture of some of the vias of an EWT silicon solar cell is shown in Figure 2.6.
CHAPTER 2. BACK-CONTACTED BACK-JUNCTION SILICON SOLAR CELLS

Figure 2.6: The laser-drilled vias of an emitter-wrap-through (EWT) silicon solar cell, shown together with a human hair for comparison of the dimensions. The picture is taken from an article by Aart Schoonderbeek and Andreas Ostendorf found on the technical news website "SPIE Newsroom", and printed with permission from Laser Zentrum Hannover (LZH).
(Source: "http://spie.org/x25041.xml?ArticleID=x25041").

2.2.2 The metallization-wrap-through cell design

The metallization-wrap-thorugh solar cell is the back-contact solar cell concept that most resembles the conventional front-contacted solar cell[31]. A schematic drawing of a MWT cell is shown in Figure 2.7. In this cell design the emitter and contact fingers are located at the front surface, but the busbars for interconnection are still at the rear side. The extension from the front side metal fingers to the busbars is done via laser drilled holes in the wafer. The holes are filled with metal through a standard metallization process, usually screen-printing. This facilitates the transition from a conventional cell process. As for EWT cells, the MWT solar cell design does not have the same requirement for long carrier lifetimes material
as BC-BJ cells, and can therefore be produced on conventional multicrystalline silicon wafers. At the same time, the reflection losses are reduced, and module assembly is still simplified compared to conventional front-contacted silicon solar cells.

2.3 Short historic overview of the BC-BJ solar cell

The BC-BJ silicon solar cell concept was first suggested by Schwartz and Lammer more than 30 years ago[16, 32]. It was originally envisioned as a cell design for concentrator applications. With the first version Schwartz reported an efficiency of 17% under 50 suns concentration. The emitter and base diffusions, and the respective contacts where arranged in an interdigitated finger pattern on the backside, see Figure 2.3. Other geometrical arrangements of the diffusions and contacts have been demonstrated as well. The most famous is the point contact

![Figure 2.7: Simplified, schematic drawing of a (p-type) metallization-wrap-through (MWT) silicon solar cell. The drawing is not to scale.](image-url)
cell of Swanson et al. [33] from 1984, where the backside diffusions are arranged in arrays of single points. An improved version of this cell was later demonstrated to give an efficiency of 28.4 % under 200 suns concentration[34]. The alignment of the backside diffusions, contact openings, and contacts were done by photolithography in early stages of BC-BJ silicon solar cell development. Four to six photolithography steps were used to produce the cells[35]. Photolithography is not compatible with cost-efficient production of solar cells, and thus the interest in the BC-BJ solar cell concept was mainly for niche and academic reasons[36].

The main contributions in the field of BC-BJ solar cells during the 1980’s and 1990’s came from Professor Swanson’s photovoltaic group at Stanford University. R. A. Sinton (also from the Stanford group) simplified the processing of BC-BJ solar cells by introducing processes for self-aligned diffusions and contact separation[35, 37]. This was a necessary step towards cost-effective BC-BJ silicon solar cells. Professor Swanson commercialized his technological ideas through the Sunpower Corporation, which he started up in 1985. Today, Sunpower is the world leading manufacturer of BC-BJ silicon solar cells, with a total production capacity of 625 MWp. In 2002, the company introduced their A-300 cell concept for one-sun solar module applications[38]. The cell is produced without any photolithography steps, and only uses low-cost manufacturing process steps. In the first round this led to an efficiency decrease compared to previous cells processed with photolithography, but the A-300 cell design has later been improved to demonstrate world-record efficiencies for BC-BJ silicon solar cells under one-sun illumination. Recently, the R&D unit at Sunpower has reported a full cell area efficiency of 23.4 % for the A-300 solar cell[18], and a large area efficiency of 20.4 % for the A-300 solar module[19], see Figure 2.4. The latter is the world record for a commercially available silicon solar module[23].

Other groups and companies are working with BC-BJ silicon solar cells as well. One very elegant cell concept that deserves to be mentioned is the RISE (Rear Interdigitated Single Evaporation) silicon solar cell from Institut für Solarenergieforschung Hameln (ISFH), developed by Engelhart et al.[39]. The RISE solar cell has been developed both as a BC-BJ silicon solar cell and as an EWT cell, see subsection 2.2.1. The cell design utilizes laser as the central processing equipment, both as a
2.4 **N-TYPE VS. P-TYPE WAFERS**

The standard in the solar cell industry has been p-type wafers with n-type diffused emitters. In theory, electrons have a higher mobility than holes, and for equal carrier lifetimes p-type wafers will therefore outperform n-type wafers due to the
CHAPTER 2. BACK-CONTACTED BACK-JUNCTION SILICON SOLAR CELLS

longer diffusion length of electrons as minority carriers. The world record 25% efficient silicon solar cell was made on p-type FZ high-purity silicon material[22]. These cells are processed in very clean facilities. In large-scale production, however, it is more or less impossible not to induce chemical and mechanical imperfections to the wafers. And herein lies the big advantage of n-type silicon solar cells. The performance of n-type cells has been shown to be less hampered by chemical and mechanical impurities. In fact, n-type silicon solar cells hold numerous advantages compared to p-type cells.

- n-type cells are not prone to light-induced degradation due to splitting of boron-oxygen complexes[44, 45].
- n-type cells are less sensitive to chemical impurities activated during high-temperature diffusion and oxidation steps[46, 47].
- n-type cells are less sensitive to laser-induced damage[45, 48].

Sunpower and Sanyo, the two companies that have reported the highest efficiencies of silicon solar cells in large-scale production both use n-type monocrystalline wafers as a base material in their A-300 and HIT solar cells, respectively[17, 49].
Chapter 3

Experimental Background

This chapter gives a general overview and a short theoretical background of the methods, materials, and equipment most frequently used during the work with this thesis.

3.1 Wafers, cleaning, and etching

3.1.1 Wafers and cleaning

We have used 4-inch, n-type, Cz, 1-3 $\Omega$-cm double-side polished wafers from Siltronix in all major experiments. The wafers have a thickness of 300 $\mu$m. p-type wafers from the same manufacturer were in some cases added to monitor the sheet resistance of phosphorus emitters. Prior to all experiments the wafers were cleaned in 5 % HF, Piranha (4:1, $\text{H}_2\text{SO}_4$: $\text{H}_2\text{O}_2$), and HCl[50].

3.1.2 Alkaline etching

After laser ablation of SiO$_2$ (see paper II), we used KOH to etch away the thermally induced damage from the laser and to reestablish a clean and smooth silicon surface. Alkaline etching of silicon and its passivating layers was extensively investigated by Seidel et al., both for undoped and doped silicon[51, 52]. In sum-
mary, they suggest the following reaction for alkaline etching of Si;

\[
\text{Si} + 2\text{OH}^- + 2\text{H}_2\text{O} \rightarrow \text{SiO}_2(\text{OH})_2^- + 2\text{H}_2, \tag{3.1}
\]

and

\[
\text{SiO}_2 + 2\text{OH}^- \rightarrow \text{SiO}_2(\text{OH})_2^- \tag{3.2}
\]

for etching of SiO₂, indicating that the alkali atoms have no effective role in the etching process. The oxide we have used during the work with this thesis is deposited at low temperature with PECVD (see section 3.2). The PECVD oxide structure is less dense than a thermal oxide, and it typically contains 1 to 10 atomic % of hydrogen[53]. The etch rate in KOH therefore differs from that reported by Seidel et al. for thermal oxides. In general, the low temperature SiO₂ does not withstand alkaline etching as well as a thermal oxide, but the etch rate is still significantly lower than for silicon.

To be able to keep a stable and controlled temperature in the aqueous KOH solution, the KOH-containing beaker was put into a larger beaker filled with water and indirectly heated by a heating plate, see Figure 3.1.

Noor et al. investigated the resulting surface roughness on n-type <100> silicon wafers after etching in aqueous KOH solutions of varying concentrations and temperatures[54], see Figure 3.2. In addition to optimizing for surface smoothness after laser ablation, the etch rate of silicon is also important to keep in mind for reduced processing times. We found, in agreement with Noor et al., that a KOH concentration of 55 % (weight) gave the smoothest surfaces for temperatures of 65 °C and 80 °C. The etch rate of silicon in alkaline solutions is decreasing with increasing alkaline concentrations above 20 % (weight)[51]. At the same time, the etch rate increases with temperature. All considerations taken into account, a 55 % (weight) KOH concentration at 80 °C gave the best results. With these etching parameters the etch rate of <100> silicon is about 0.6 µm per minute[51], while the measured etch rate of PECVD SiO₂ was found to be approximately 10 nm per minute.

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¹Noor et al. do not explain how the surface roughness is measured, and we therefore assume that the roughness is the arithmetic average, \( R_a \), from the surface mean, which is the most common way of defining it.
3.1. WAFERS, CLEANING, AND ETCHING

Figure 3.1: The setup used to keep a stable temperature in the aqueous KOH etching solution.

Figure 3.2: The resulting surface roughness as a function of temperature on <100> n-type silicon after etching in aqueous KOH solutions of varying temperatures and concentrations (weight %). The data points are from Noor et al.[54].
3.2 PECVD

Plasma-enhanced chemical vapor deposition (PECVD) is a technique commonly used to deposit thin films from vapor phase to solid state on substrates. It is a subset of general chemical vapor deposition techniques that is aided by a plasma (normally) created by a radio-frequency discharge between two electrodes. The deposition is a result of chemical reactions in or between one or several precursor gasses in the plasma and the substrate. The volatile byproducts are removed by the gas flow through the reactor chamber. A Plasmalab 133 parallel plate system from Oxford Instruments[55] was used during the work with this thesis, see Figure 3.3. The substrate is placed on the lower electrode, which is heated. The combination of different precursor gases, deposition time, temperature, pressure, power density, and flow of the different gases make it possible to deposit a wide range of thin films. Each type of film can be optimized for its purpose by altering the deposition parameters and by finding the optimum ratio between the flow of different precursor gases. In this work the Oxford Instruments PECVD system has been used to deposit hydrogenated amorphous silicon (a-Si:H, a-Si for simplicity) and hydrogenated amorphous silicon oxide (a-SiO$_x$:H, SiO$_2$ for simplicity).

3.2.1 Hydrogenated amorphous silicon

Crystalline silicon (c-Si) has a diamond like structure based on covalent sp$^3$ bonding. The convenient electrical and optical properties of c-Si, in combination with its abundance, have made it the most commonly used material in the semiconductor industry. Amorphous silicon, on the other hand, is a disordered form of silicon, and the physical properties differ significantly from that of c-Si. In contrast to c-Si, a-Si has a direct optical bandgap of approximately 1.7 eV[56]. It is, thus, a strong optical absorber in the visible part of the spectrum. The disordered atomic structure also implies that a-Si has a low thermal conductivity. The conductivity depends on the hydrogen content, but is in the order of 0.01 W/cm-K[57, 58]. This is more than two orders of magnitude lower than that of c-Si, see Figure 3.10.
In addition, there has been shown to be an additional thermal resistance at the interface between thin film a-Si and bulk c-Si. This was investigated and quantified by Kuo et al. [59] and Volz et al.[60], and will be discussed more thoroughly in sections 3.3.1 and 3.3.2.

Locally, a-Si resembles its crystalline counterpart, but there is no long range order. The atomic structure of a-Si:H as obtained from density function theory (DFT) calculations is shown in Figure 3.4. Each Si atom is in general bounded to four neighbors at tetrahedral angles ($\theta = 109^\circ$) and a bond length of 0.35 nm. A spread in bond angle of about 10 % and a spread in bond length of approximately 2 % cause the disordering. a-Si also contains a large portion of dangling (unsatisfied) bonds. The density of these imperfections is typically in the order of $10^{15}$ to $10^{16}$ cm$^{-3}$. When prepared by PECVD, a-Si contains in the order of 10 atomic % of...
CHAPTER 3. EXPERIMENTAL BACKGROUND

Figure 3.4: The atomic disordered structure of a-Si:H as obtained by density functional theory (DFT) calculations with the Vienna Ab-Inito Simulation Package (VASP)[61, 62, 63, 64]. The grey and black balls represent silicon and hydrogen atoms, respectively. The hydrogen is found both interstitially and bonded to silicon dangling bonds. The simulations are performed by Smagul Karazhanov at the Department of Solar Energy, IFE.

Hydrogen[65]. Hydrogen is known to passivate dangling bonds, and thereby lower the impact of the defects in the amorphous structure.

For deposition we used undiluted silane (SiH₄) as the only precursor gas in the PECVD chamber[66]. The deposition temperature and pressure were 230 °C and 300 mTorr, respectively. When a-Si was used as a surface passivation layer, a very
low RF input power density of 0.008 W/cm² was used during deposition to avoid surface damage as a consequence of ion bombardment. A slightly higher power density was used to ignite the plasma during the first few seconds of the deposition. With this power density we measured the deposition rate to be approximately 5 nm per minute. The overall reaction that must occur is

\[
\text{SiH}_4(g) \rightarrow \text{Si}(s) + 2\text{H}_2(g),
\]

where s and g denote solid and gas phase, respectively. In general, the reaction is usually more complicated. The oscillating electric field causes the unstable silane to dissociate, producing a mixture of ions, electrons, and radicals; a plasma. The dominant precursor to a-Si:H is the SiH₃ radical. This radical, which is short of one hydrogen atom, and so has a reactive dangling bond, is transported to the substrate by the gas flow. When SiH₃ attaches to the surface where there is a dangling bond, this will lead to film growth[67].

The implementation of hydrogenated amorphous silicon to the solar cell and thin film transistor (TFT) technology has been challenged by the well described metastability of the material. Both light and applied bias induce chemical changes in a-Si[67]:

\[
\text{SiHHSi + Si–Si} \leftrightarrow 2\text{SiHDSi},
\]

where Si–Si represents a weak bond, and D represents a dangling bond. For TFTs, the increased number of dangling bonds, i.e., bandgap states, leads to a shift in the threshold voltage[67], while for a-Si solar cells it induces a degradation of the photocurrent[68]. The latter is known as the Staebler-Wronski effect[69, 70]. The induced changes have been shown to be reversible upon annealing. Since the light-induced carriers break Si–H bonds, it also degrades the surface passivation quality of a-Si films on Si. For backside passivation, however, this is less of an issue since the charge carrier generation in the a-Si layer is very low. The passivation quality of backside a-Si has been reported to be long-term stable[71, 72].

In contrast to e.g. PECVD SiNx, the passivation mechanisms in a-Si on Si are purely chemical[71, 73]. The passivation originates from termination of dangling
bonds at the interface and not by an electric field induced by surface charge. This is important in BC-BJ silicon solar cells to avoid channel tunneling near the back-surface[74]. A capacitance-voltage ($C-V$) plot of an a-Si film on a c-Si substrate is shown in Figure 3.5. The clean exponential form of the curves indicates that there is no surface charge between the materials, and that we have a (near) perfect surface between the amorphous film and the crystalline bulk. Finally, a-Si has been shown to be excellent for passivation of c-Si surfaces, either solely or in a stack with a PECVD dielectric[65, 71, 72, 75, 76, 77, 78].

We have used the qualities of a-Si for several purposes during the work in this thesis. In paper II a thin layer of a-Si was deposited on both sides of each sample for surface passivation, i.e. to terminate the dangling bonds at the c-Si wafer surface. This was done after laser processing, etching, and cleaning and prior to lifetime measurements. Surface passivation is done to, as far as possible, exclude surface recombination from the effective recombination lifetime measurements, see section 3.5. In paper III other aspects of the a-Si film is exploited. We took advantage of the high optical absorption and low thermal conductivity of the material. a-Si is utilized to absorb laser irradiation and to locally remove a covering transparent

![Figure 3.5: Capacitance-voltage ($C-V$) plot of a thin film of a-Si on n-type bulk c-Si at different AC frequencies.](image-url)
layer of PECVD SiO₂ for point contact formation. The high optical absorption and low thermal conductivity of a-Si, aided by the interface resistance to c-Si, induce a high energy density in the a-Si film after laser irradiation. This leads to an indirect ablation of SiO₂ at laser fluences (energy densities) five times lower than that needed to ablate SiO₂ on c-Si. The residual SiO₂–a-Si double-layer has been shown to be excellent for surface passivation for the back of a silicon solar cell[71, 72]. With the modeling work in paper IV we aimed to get a better theoretical understanding of the physics involved in laser ablation of SiO₂ with and without the buffer layer of a-Si. Both a static and a dynamic model were developed, and both were in good agreement with the experimental data, see sections 3.3.1 and 3.3.2. In paper V we investigated the contact resistance for contacting through the same passivating double-layer of a-Si and SiO₂.

### 3.2.2 Hydrogenated amorphous silicon oxide

SiO₂ (a-SiO₂:H) was deposited with the Oxford Instruments Plasmalab PECVD system described above. To get deposition we must have;

\[
\text{SiH}_4(g) + 2\text{O} \rightarrow \text{SiO}_2(s) + 2\text{H}_2. \tag{3.5}
\]

O₂ can be used, but since oxygen is reactive with silane independent of the plasma, considerable nucleation will take place uniformly in the whole PECVD system, not only on the sample. Thus, other oxygen sources are usually preferred. CO₂ is also a likely source of oxygen, but carbon incorporation can have a deleterious effect on device performance. N₂O is therefore usually preferred in PECVD oxide deposition processes. The probable chemical route is

\[
3\text{SiH}_4 + 6\text{N}_2\text{O} \rightarrow 3\text{SiO}_2 + 4\text{NH}_3 + 4\text{N}_2. \tag{3.6}
\]

We used a flow of 1000, 800, and 17 sccm of N₂O, N₂, and SiH₄, respectively. The temperature, pressure, and RF input power density were 300 °C, 500 mTorr, and 0.06 W/cm², respectively. The flow of N₂ improves the homogeneity of the oxide film as it dilutes the other precursor gasses, but it does not take part in the chemical reactions. We found the deposition rate to be approximately 45 nm per minute.
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Since the PECVD oxide is deposited at low temperature, and not thermally grown, it has an amorphous structure. The atomic structure of a-SiO$_2$ as obtained from DFT calculations is shown in Figure 3.6. It is less dense than a thermal oxide, and PECVD SiO$_2$ typically has a hydrogen content of 1-10 atomic% [53]. It has been used for several purposes during the work with this thesis.

- It has been used as a diffusion barrier against liquid dopant-containing spray-on sources, and against gaseous BBr$_3$ and POCl$_3$ in our tube furnace. The SiO$_2$ has a finite resistance against dopants, and a thick layer was needed to get sufficient resistance against the dopants for typical diffusion processes, see section 4.1.

- The oxide was used as an etching barrier against aqueous KOH solutions. KOH was used to clean up the induced damage after laser ablation of large areas (see paper II). In contrast to a-SiN$_x$, amorphous silicon oxide has a finite etch rate in KOH. This etch rate, which depends on the KOH concentration and temperature, needs to be taken into account under device fabrication. It is important because a certain thickness of the oxide film is needed to make it work successfully as a diffusion barrier. In addition, laser parameters used for structuring (see paper II) and contact opening (see paper III and V) are sensitive to the thickness of the oxide layer due to interference effects, see Figure 3.7. The etch rate of PECVD SiO$_2$ in an aqueous KOH solution of 55 % (weight) at 80 °C was found to be less than 10 nm per minute.

- The oxide served as an anti-reflection coating for the green laser light during indirect ablation of oxide, see section 3.3. An optimized thickness of the oxide layer lowers the energy threshold for ablation. This is done by tailoring the thickness of the SiO$_2$ into a reddish color, which implies that it will have low reflection in the green part of the spectrum, see Figure 3.7.
3.2. PECVD

Figure 3.6: Atomic structure of a-SiO$_2$ as obtained by density functional theory (DFT) calculations with the Vienna Ab-Initio Simulation Package (VASP)[61, 62, 63, 64]. The grey and red balls represent silicon and oxygen atoms, respectively. The simulations were performed by Smagul Karazhanov at the Department of Solar Energy, IFE.
Figure 3.7: Reflectance measurements of a 450 nm thick SiO$_2$ film on c-Si bulk. The reflection increases from 18 % as-deposited and near optimized (solid black line) to 37 % after immersing the sample to an aqueous KOH etching bath (grey dotted line) for a wavelength of 532 nm (indicated by the vertical dotted line), corresponding to that of the green laser used in these experiments.
3.3 Laser

The laser we have used during the work with this thesis is a diode pumped, frequency doubled (532 nm), Nd:YVO$_4$ RSM 20E from Rofin-Sinar. The system is Q-switched, and the pulse duration varies between 30 and 350 nanoseconds depending on the diode current and the repetition frequency. In focus the $\frac{1}{e^2}$ spot diameter is 40 $\mu$m. The laser beam characteristics were measured by a BC106-VIS CCD camera beam profiler from Thorlabs. With this laser system we have the possibility of varying the diode current from 12 A to 40 A, the pulse repetition frequency from 10 kHz to 100 kHz, in addition to changing the focal plane. The other laser parameters, such as pulse duration and energy, vary internally with the diode current and pulse frequency. The measured fluence as a function of diode current at 20 kHz repetition frequency is shown in Figure 3.8. The two different curves in the figure represent measurements of the laser fluence with six months time difference with the same parameter settings. The difference was found to result from a degradation of the laser diode power. We measured the fluence with an Ophir Laserstar power meter and a BFi-OPTiLAS pyroelectric sensor.

The laser has been a central processing tool during the work with all the experimental papers (II, III, and V). We have used the same laser to structure large areas for patterning of the backside of BC-BJ silicon solar cells (see paper II), and to make local contact openings through a stack of PECVD a-Si and SiO$_2$. The latter was investigated in papers III, IV, and V. First through the influence on the local surface recombination velocity (see subsection 3.5.1), then by modeling (see subsections 3.3.1 and 3.3.2), and finally by investigating the contact resistance by using the transfer length method (see section 3.7).

3.3.1 Laser interaction with crystalline and amorphous silicon

In both crystalline and amorphous silicon, laser energy is absorbed by excitation of charge carriers across the bandgap. The optical (linear) absorption coefficient, $\alpha$, is given as

$$\alpha = \frac{4\pi k}{\lambda}, \quad (3.7)$$
where $\lambda$ denotes the wavelength, and $k$ the imaginary part of the complex refractive index, $\tilde{n} = n - ik$. The optical intensity at a depth $x$ in the material can be calculated by the Bouger-Beer-Lambert law through[79]

$$\frac{dI(x)}{dx} = -\alpha x. \tag{3.8}$$

A possible solution to this equation for an incident optical intensity $I_0$ at a sample with surface reflection $R$ is

$$I = I_0(1 - R)e^{-\alpha x}, \tag{3.9}$$

if we assume that the attenuation of the optical intensity is solely caused by absorption. The $\frac{1}{c}$ optical absorption depth is then simply given by $\frac{1}{\alpha}$, and often denoted $l_\alpha$. In the most general case, $\alpha$ is temperature dependent. The temperature dependent optical absorption coefficient of silicon, $\alpha(T)$, at $\lambda = 532$ nm has

---

2The law is also frequently called the Lambert-Beer law or simply Beer’s law. It was first derived experimentally by P. Bouger in 1729 and then theoretically by I. G. Lambert in 1760. A mesoscopic interpretation of $\alpha$ was first given by A. Beer in 1952.

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been measured and parameterized by Jellison et al.[80];

\[ \alpha(T) = 5.02 \times 10^3 e^{T/430}, \]  

(3.10)

for \(300 \text{ K} < T < 1000 \text{ K}\), see Figure 3.9. The referenced temperature range is marked with a double arrow in the figure. In the simulations we have continued this exponential increase also in the range 1000 K to 1685 K, where the latter corresponds to the melting temperature of c-Si. At the transition temperature an abrupt change in the absorption coefficient is marked in the figure (3.9)[81]. For liquid silicon \((T > 1685 \text{ K})\), an absorption coefficient of \(1 \times 10^6 \text{ cm}^{-1}\) (still for \(\lambda = 532 \text{ nm}\)) was measured by Rezek et al.[81]. For a-Si, we do not have data other than for at 300 and 1420 K, marked with circles in the figure. We know, however, that the absorption coefficient of a-Si has been found to increase linearly with temperature at other wavelengths[82], and we assume this to be valid also at 532 nm.

The first interaction between laser and material is always purely optical, but when the duration of the laser pulse exceeds the duration of the thermal relaxation mechanisms in the material (in the order of \(10^{-11} \text{ s}\) in silicon), heat will start to dissipate into the material. The absorbed laser energy is transferred via electron-electron, electron-phonon, and phonon-phonon interactions. The thermal diffusion length, \(l_{th}\), is often approximated by

\[ l_{th} \approx \sqrt{2D\tau_p}, \]  

(3.11)

where \(D\) is the thermal diffusivity and \(\tau_p\) is the laser dwell time. The thermal diffusivity quantifies a material’s ability to adjust its temperature according to the surroundings. It is given as the ratio between the thermal conductivity, \(\kappa\), and the volumetric heat capacity, \(C_p = \rho c_p\), with \(\rho\) being the mass density and \(c_p\) the specific heat capacity;

\[ D = \frac{\kappa}{\rho c_p}. \]  

(3.12)
Figure 3.9: The absorption coefficients of crystalline and amorphous silicon at a wavelength of 532 nm. The referenced temperature range is marked with a double arrow on the c-Si curve. For the a-Si curve the only two referenced temperatures are marked with circles.

$l_{th}$ describes the $\frac{1}{e^{1/2}}$ spatial decay\textsuperscript{3} in the temperature distribution

$$T(x, t) \approx \frac{Q}{\rho c_p \sqrt{4\pi D t}} e^{-\frac{(x)^2}{4Dt}}.$$  

This is a fundamental solution of the one-dimensional linear heat equation for a point source in infinite space;

$$\frac{\partial T}{\partial t} = D \frac{\partial^2}{\partial x^2} T + \frac{Q}{\rho c_p} \delta(x, t),$$  

where $Q$ is the total energy release, and $\delta$ the Dirac delta function.

In general, $l_{th}$ depends on the particular boundary problem under consideration, and may differ from Equation (3.11). Both the thermal conductivity, the heat capacity, and the density of mass are usually temperature dependent. In addition, the laser source has a finite distribution in space and time. This leads to a more

\textsuperscript{3}The $\frac{1}{e}$ spatial decay in the temperature is also often used to describe $l_{th}$, which then can be approximated as $2\sqrt{Dt/\rho}$. 

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Figure 3.10: The thermal conductivity of c-Si and a-Si:H. The values for the solid state c-Si and a-Si have been imported from Ong et al.[58], while the thermal conductivity of liquid silicon was reported by Kobatake et al.[83]. We have simulated the thermal conductivity with abrupt changes at the transition temperature.

The general version of the heat diffusion equation (still in one-dimension)\(^4\):

\[
c_p(T)\rho(T)\frac{\partial T(x,t)}{\partial t} = \frac{\partial}{\partial x} \cdot \kappa(T) \left( \frac{\partial}{\partial x} \nabla \cdot T(x,t) \right) + S(x,t) \tag{3.15}
\]

where the laser source term, \(S(x,t)\), is given by

\[
S(x,t) = \sqrt{\frac{4 \ln 2}{\pi}} \frac{(1 - R)F_p \cdot \alpha}{\tau_p} \exp \left( -\alpha x - 4 \ln 2 \frac{(t - t_{\text{peak}})^2}{\tau_p^2} \right), \tag{3.16}
\]

and has units of W/cm\(^3\). \(t_{\text{peak}}\) denotes the time for the peak fluence, \(F_p\), in the Gaussian time distribution. The pulse duration, \(\tau_p\), is given at full-width-half-maximum (FWHM). The fluence is the pulse energy density, and it is given in J/cm\(^2\). The peak fluence, \(F_p\), of a perfect Gaussian beam profile is two times the average energy density:

\[
F_p = \frac{2E_p}{\pi \phi_0^2}, \tag{3.17}
\]

where \(E_p\) is the pulse energy, and \(\phi_0\) is the \(\frac{1}{e^2}\) radius of the Gaussian beam. In

\(^4\)This equation does not take into account heat transport by convection and thermal radiation.
the modeling we have limited ourselves to one-dimensional simulations, and have therefore only used the Gaussian peak.

The complexity of Equation (3.15) is now far beyond the analytically solvable. We therefore applied a finite difference scheme to solve the equation. To also include the possibility of phase transitions we upgraded to the so-called total enthalpy version of Equation (3.15)[79];

$$\frac{\partial \Delta H(x, t)}{\partial t} = \frac{\partial}{\partial x} \cdot \kappa(T) \left( \frac{\partial}{\partial x} \nabla \cdot T(x, t) \right) + S(x, t), \quad (3.18)$$

where $\Delta H(x, t)$ is the total enthalpy;

$$\Delta H(x, t) \approx \int_{T_i}^{T} \rho(T')c_p(T')dT' + \Theta(T - T_m)\Delta H_m + \Theta(T - T_v)\Delta H_v. \quad (3.19)$$

$\Theta$ is the Heaviside function (1 if argument is positive, 0 if negative), $\Delta H_m$ and $\Delta H_v$ are the enthalpies of melting and vaporization, respectively, with $T_m$ and $T_v$ being the corresponding melting and vaporization temperatures. Since the phase transitions are of first order their derivatives become infinite at the transition temperature. This can potentially lead to numerical instabilities when solving Equation (3.18). We avoided the instabilities by representing the Dirac delta function by a Gaussian function with $\Delta T = 5$ K.

The sample is assumed to be at ambient temperature before each laser pulse. This assumption gives us the initial condition;

$$T(x, t) = T_0|_{t=0}. \quad (3.20)$$

Our experiments are performed at 20 kHz laser repetition frequency. This implies that there are 50 $\mu$s between each pulse which has a duration of approximately 100 ns. The initial condition is therefore valid also for spatial overlapping pulses since the time between two pulses is significantly longer than it takes for the sample to be back to ambient temperature after the pulse. We also assume that the
temperature is equal to the ambient temperature at infinite depth in the material;

\[ T(x, t) = T_0|_{x \to \infty}. \] (3.21)

We have run the simulations with 10 \( \mu m \) of c-Si under the a-Si film. This is sufficient to keep the error associated with boundary condition 3.21 at a negligible level. Finally, we have also assumed that there is no energy loss at the front surface;

\[ \frac{\partial T}{\partial x} = 0|_{x=0}. \] (3.22)

This is a simplification, but for pulsed laser sources the energy loss at the surface, usually due to emission and convection, is negligible compared to the energy diffusion towards the bulk of the c-Si wafer. The left hand side of Equation (3.22) is therefore usually put equal to zero with a good approximation[84].

The interface thermal resistance between a-Si and c-Si was implemented by introducing the resistance over a layer with a thickness of 5 nm between the materials. A value of 0.0054 cm\(^2\)·K/W was used according to Kuo et al. [59], corresponding to a thin layer of native oxide between a-Si and c-Si. In general, the interface between two different materials is always resistive to heat flow, even for atomically perfect interfaces. This is due the different vibrational and electronic properties on each side of the interface. The interface causes scattering of both electrons and phonons, and it leads to a discontinuity in the temperature distribution across the surface.

### 3.3.2 Laser fluence ablation threshold

The dynamic model described in the previous subsection gives insight into the temporal and one-dimensional spatial temperature distribution of heat in a-Si on c-Si. We found a good agreement between experimental and modeled data. In order to be able to predict the laser fluence ablation threshold for a given thickness of the a-Si film, we have also developed a simplified analytically solvable static laser ablation model. The model is based on a model described by Matthias et al.[85]
CHAPTER 3. EXPERIMENTAL BACKGROUND

when studying laser ablation of metal films on quartz. Our version of the model has been revised to also include the possibly of optical absorption in the substrate, in our case c-Si, and a thermal diffusion barrier at the a-Si–c-Si interface. All parameters are assumed to be constant, and we have used the values at 1000 K listed and referenced in paper IV. We assume that the absorbed laser irradiation is completely transferred into heat, and that there is a uniform temperature rise within the volume defined by the thermal diffusion length, \(l_{th}\), after the laser pulse. By the first law of thermodynamics

\[
\Delta T = \frac{Q_{abs}}{C_p}. \quad (3.23)
\]

\(\Delta T\) is the temperature increment, \(Q_{abs}\) is the absorbed laser energy (density) and \(C_p\) is the volumetric heat capacity at constant pressure. We assume that a phase transition (solid-liquid or liquid-vapor) occurs at a critical energy density, \(\epsilon_T = F_T/l_{th}\), where \(F_T\) is the laser fluence threshold at the critical energy density. By accounting for the reflection, \(R\), at the surface and light that is not absorbed in the film, \((1-e^{-\alpha d})\), nor in the substrate, \((1-e^{-\beta d_s})\), where \(\alpha\) and \(\beta\) and \(d\) and \(d_s\) are the optical absorption coefficients and thicknesses of the a-Si film and the c-Si substrate, respectively, we end up with the following expression for the absorbed energy:

\[
Q_{abs} = F \cdot l_{th} (1 - R) \left[ (1 - e^{-\alpha d}) + e^{-\alpha d} (1 + e^{-\beta d_s}) \right], \quad (3.24)
\]

where \(F\) is the incident pulse fluence. By setting \(Q_{abs}\) equal to the critical energy density, \(\epsilon_T\), at the temperature of the phase transition, and combining Equations (3.23) and (3.24), we get the following expression for the fluence threshold for melting

\[
F_{TM} = \frac{\Delta T_M}{(1 - R) \left[ (1 - e^{-\alpha d}) + e^{-\alpha d} (1 + e^{-\beta d_s}) \right]} l_{th} C_p, \quad (3.25)
\]

where \(\Delta T_M\) is the needed temperature increment for melting. Still, the above equation does not take into account the different thermal diffusion lengths and the different heat capacities of a-Si and c-Si. In the following the heat capacities have been taken additively into account and weighted between the film and the substrate. It changes linearly from the heat capacity of c-Si at \(d = 0\) (surface) to the heat capacity of a-Si at \(d = l_{th}\) and above. We define this heat capacity as the
3.3. LASER

effective heat capacity, $C_{p,\text{eff}}$. We also need to find an effective thermal diffusion length, $l_{\text{th,eff}}$. In order to find an appropriate expression for the latter, we need to find an effective thermal conductivity, and thereby an effective diffusivity, see equation (3.11). The effective thermal conductivity is found by adding the thermal resistivities of a-Si, c-Si, and the interface as

$$\frac{d + l_{\text{th,si}}}{\kappa_{\text{total}}} = \frac{d}{\kappa_{\text{aSi}}} + \frac{l_{\text{th,si}}}{\kappa_{\text{cSi}}} + R_{\text{int}},$$  

(3.26)

where $\kappa_{\text{aSi}}$ and $\kappa_{\text{cSi}}$ are the thermal conductivities of a-Si and c-Si, respectively, $R_{\text{int}}$ is the interface resistance, and $l_{\text{th,si}}$ is the thermal diffusion length of c-Si. Finally,

$$D_{\text{eff}} = \frac{\kappa_{\text{total}}}{C_{p,\text{eff}}}.$$

(3.27)

and $l_{\text{th,eff}} = \sqrt{2D_{\text{eff}}\tau_{\text{p}}}$. This leaves us with the following expression for the fluence ablation threshold for melting:

$$F_{TM} = \frac{\Delta T_{M}}{(1 - R) [(1 - e^{-\alpha d}) + e^{-\alpha d}(1 - e^{-\beta d})]} l_{\text{th,eff}} C_{p,\text{eff}}.$$

(3.28)

The corresponding expression for surface vaporization can then easily be found by replacing $\Delta T_{M}$ by the needed temperature increment for vaporization, $\Delta T_{V}$, plus the latent heat for melting divided by the volumetric heat capacity of a-Si;

$$F_{TV} = \frac{\Delta T_{V} + \Delta H_{M}/C_{p,\text{aSi}}}{(1 - R) [(1 - e^{-\alpha d}) + e^{-\alpha d}(1 - e^{-\beta d})]} l_{\text{th,eff}} C_{p,\text{eff}}.$$

(3.29)

The latter model was found to be in good agreement with experimental data for the fluence threshold for laser ablation when including a thermal interface resistance of 0.0054 cm$^2$·K/W, which corresponds to a sample with a native oxide between a-Si and c-Si[59, 60]. The match between the experimental data from paper III and the model and the effect of altering the interface thermal resistance from 0.0013 cm$^2$·K/W to 0.0054 cm$^2$·K/W can be seen in Figure 3.11. 0.0013 cm$^2$ corresponds to a clean surface between the two materials[59]. The covering PECVD oxide is not regarded as a part of the thermal simulations. It has an effect on the surface reflection, which has been taken into account, but
CHAPTER 3. EXPERIMENTAL BACKGROUND

the oxide is regarded thermally insulating.
Figure 3.11: (a) Experimental data for the laser fluence ablation threshold of a-Si as a function of the film thickness (solid line with circles), shown together with the models for surface melting (Equation (3.28)) and vaporization (Equation (3.29)). The experimental data is from paper III. (b) The effect of increasing the interface thermal resistance from 0.0013 to 0.0054 cm²·K/W in the model for surface vaporization (Equation (3.29)).
3.4 Doping

During the work with this thesis, the facilities at the IFE solar laboratory have been gradually upgraded. Initially, doping of wafers was done by a custom-made spray-on system where diluted spin-on-dopant (SOD) sources were sprayed on to the samples. SODs typically consist of phosphorous pentoxide ($\text{P}_2\text{O}_5$) or boron trioxide ($\text{B}_2\text{O}_3$) interchained with SiO$_2$ dissolved in alcohols and ketones[86]. The samples are dried to remove the solvents. Phosphorus and boron are then released upon reduction of $\text{P}_2\text{O}_5$ and $\text{B}_2\text{O}_3$, respectively, at high temperature in a belt furnace as:

$$2\text{P}_2\text{O}_5(l) + 5\text{Si}(s) \rightarrow 4\text{P}(s) + 5\text{SiO}_2(s), \quad (3.30)$$

and

$$2\text{B}_2\text{O}_3(l) + 3\text{Si}(s) \rightarrow 4\text{B}(s) + 3\text{SiO}_2(s), \quad (3.31)$$

and driven into the substrate.

Figure 3.12 shows a spatially resolved sheet resistance map of a series of wafers used in a preliminary experiment to optimize the homogeneity of boron-emitters on n-type 5 inch Cz wafers. The external air flow of the atomizing nozzle is gradually reduced from left to right. Though the optimized result shown in the figure is satisfactory, the spray-on system was found hard to maintain and keep stable, and re-optimization of the spraying parameters had to be done frequently. We therefore switched to using a tube furnace when this became available in the IFE laboratory.

The diffusion process in a tube furnace is fundamentally the same as diffusion in a belt furnace. It can also be divided into two sequences; deposition and drive-in. In contrast to SODs, where $\text{P}_2\text{O}_5$ and $\text{B}_2\text{O}_3$ and are sprayed directly onto the silicon wafer, gaseous phosphorus oxychloride (POCl$_3$) or boron tribromide (BBr$_3$) is fed into the tube to react with externally supplied O$_2$ as

$$4\text{POCl}_3(g) + 3\text{O}_2(g) \rightarrow 2\text{P}_2\text{O}_5(l) + 6\text{Cl}_2(g), \quad (3.32)$$
3.5. LIFETIME AND SATURATION CURRENT MEASUREMENTS

and

\[ 4\text{BBBr}_3(g) + 3\text{O}_2(g) \rightarrow \text{B}_2\text{O}_3(l) + 6\text{Br}_2(g), \]  

(3.33)

for phosphorous and boron diffusion, respectively\cite{86, 87}. The gaseous POCl\textsubscript{3} or BBBr\textsubscript{3} condense on the silicon wafer surface and form a solid oxide. The drive-in step is then activated by temperature, and the reduction takes place as describe above in Equations (3.30) and (3.31) for phosphorous and boron, respectively. We found the homogeneity and reproducibility of the diffused layers to be significantly improved with the tube furnace\textsuperscript{5}.

![Spatially resolved sheet resistance map of a series of wafers used for optimization of the boron- emitter homogeneity deposited by our custom-made spray-on system. The external airflow of the atomizing nozzle was reduced from left-most to the right-most picture. The resulting p\textsuperscript{+} sheet resistance for the optimized spray-on parameters in these experiments had a spread in homogeneity of < 3 % at 60 \(\Omega\)/\(\square\).](image)

Figure 3.12: Spatially resolved sheet resistance map of a series of wafers used for optimization of the boron- emitter homogeneity deposited by our custom-made spray-on system. The external airflow of the atomizing nozzle was reduced from left-most to the right-most picture. The resulting p\textsuperscript{+} sheet resistance for the optimized spray-on parameters in these experiments had a spread in homogeneity of < 3 % at 60 \(\Omega\)/\(\square\).

### 3.5 Lifetime and saturation current measurements

The effective minority carrier lifetime, \(\tau_{\text{eff}}\), of a silicon wafer can be expressed as the sum of the recombination rates in the bulk and on the surface. The recombination rate is in general inversely related to the recombination lifetime, and the

\textsuperscript{5}The greatest disadvantage of POCl\textsubscript{3} and BBBr\textsubscript{3} diffusion is the batch nature of the process. It is time and energy consuming, and as the industry is heading towards thinner wafers, the common way a stacking wafers vertically in a quartz boat will eventually leave manufacturers with an unacceptable yield\cite{86}.  

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effective lifetime of a sample can be expressed as

\[
\frac{1}{\tau_{\text{eff}}} = \frac{1}{\tau_{\text{bulk}}} + \frac{1}{\tau_{\text{surface}}},
\]

(3.34)

where \(\tau_{\text{bulk}}\) and \(\tau_{\text{surface}}\) are the bulk and surface recombination lifetimes, respectively. The recombination rate in the bulk can further be divided into the sum of independent recombination mechanisms via the inverse recombination lifetimes as

\[
\frac{1}{\tau_{\text{bulk}}} = \frac{1}{\tau_{\text{rad}}} + \frac{1}{\tau_{\text{Auger}}} + \frac{1}{\tau_{\text{SRH}}},
\]

(3.35)

\(\tau_{\text{rad}}, \tau_{\text{Auger}},\) and \(\tau_{\text{SRH}}\) are the radiative, Auger, and Shockley-Read-Hall recombination lifetimes, respectively. Since silicon is an indirect bandgap semiconductor, radiative recombination is low and negligible to a good approximation. Auger recombination is intrinsic to the material, and depends only on the base resistivity. It therefore sets an upper boundary for the theoretical effective lifetime of silicon. Shockley-Read-Hall recombination, on the other hand, occurs via defects levels in the bandgap, and depends therefore on the material quality. It is usually the dominant bulk recombination source in solar cells.

The surface of a silicon wafer can be regarded as a large imperfection in the crystal, and it represents a continuum of defects in the bandgap. If the surface recombination rate is low and equal on each side, \(\tau_{\text{surface}}\) can be approximated by

\[
\frac{1}{\tau_{\text{surface}}} = \frac{2S_{\text{eff}}}{W},
\]

(3.36)

where \(W\) is the wafer thickness and \(S_{\text{eff}}\) is the surface recombination velocity.

In silicon solar cells recombination in the emitter is also usually significant. It is therefore common to add the emitter recombination rate, \(\frac{1}{\tau_{e}}\), to the other bulk recombination mechanisms in the expression for the effective recombination lifetime (see Equation (3.35)). Under the assumptions that the emitter is in low
3.5. LIFETIME AND SATURATION CURRENT MEASUREMENTS

... and that the depletion approximation is valid, it is possible to get an analytical expression for an effective virtual surface recombination velocity that includes both surface and emitter recombination. The emitter, or depletion region, is treated as a virtual surface, and it can be shown that expression for the effective emitter recombination velocity is given as [78, 86]

\[ S_{\text{eff}} = J_{0e} \frac{N_D + \Delta p}{q n_i^2}, \]  

(3.37)

where \( J_{0e} \) is the emitter recombination current, \( N_D \) is the base doping density of an n-type wafer, \( n_i \) is the intrinsic carrier density, and \( \Delta p \) is the excess/photogenerated carrier density. By inserting equations (3.36) and (3.37) into 3.35, the total effective recombination lifetime of a silicon wafer with two equal emitters can be expressed as [88, 89]

\[ \frac{1}{\tau_{\text{eff}}} = \frac{1}{\tau_{\text{SRH}}} + \frac{1}{\tau_{\text{auger}}} + 2J_{0e} \frac{N_D + \Delta p}{W q n_i^2}. \]  

(3.38)

In the next section we will show that the latter expression can be used to extract the empirical value for the emitter saturation current, \( J_{0e} \), as long as the recombination lifetime is not limited by Auger recombination in the bulk [89].

3.5.1 The generalized photoconductance method

The generalized photoconductance method combines the steady state and transient photoconductance methods into a unified robust measurement technique [90, 91]. The carrier continuity equation can be expressed as

\[ \frac{d\Delta n}{dt} = g_E - U, \]  

(3.39)

when there is no net current transport in the sample. \( \Delta n \) is the excess/photogenerated carrier density, and \( U \) and \( g_E \) is the recombination and generation rate, respect-
tively. By simple manipulation we find that

$$
\tau_{\text{eff}} = \frac{\Delta n}{g_E - \frac{d\Delta n}{dt}}, \quad (3.40)
$$

under the assumption that $\tau_{\text{eff}}$ is inversely proportional to the recombination rate. The charge carriers generated in the sample due to illumination cause the conductance of the sample to increase. This elevated conductance is known as the photoconductance, $\sigma_p$. Assuming that the excess carrier densities of majority and minority carriers are equal, the photoconductance can be expressed as

$$
\sigma_p = q\Delta n(\mu_n + \mu_p)W, \quad (3.41)
$$

where $\Delta n$ now is the average excess carrier density throughout the whole thickness, $W$, of the wafer. $\mu_n$ and $\mu_p$ are the electron and hole mobilities, respectively.

We used a Sinton WTC-100 setup to extract the emitter saturation current, $J_{0e}$, under high injection conditions. To get accurate results the injection of minority carriers needs to be at least 10 times the doping density[88]. Trap-assisted recombination mechanisms saturate at a fixed level when the injection of photogenerated carriers exceed the doping density. Since Auger recombination is intrinsic to the material, it can be deducted from Equation (3.38). This leaves the emitter recombination current as the only unknown and the parameter limiting the effective lifetime. $J_{0e}$ is calculated as the slope of the inverse Auger-corrected lifetime, which becomes linear at high injection. An output window from the Sinton WTC-100 software setup is shown in Figure 3.13.

### 3.5.2 $\mu$W-PCD

In microwave PCD ($\mu$W-PCD) the photoconductance is measured through the reflectance of a microwave signal directed at the sample. The decay of the reflected microwave power is utilized for estimation of the excess carrier decay. A laser pulse is commonly used to generate the charge carriers, since the very rapid decay times of laser diodes ensure negligible steady state contributions even for low
3.5. LIFETIME AND SATURATION CURRENT MEASUREMENTS

Figure 3.13: Screen print from the Sinton WTC-100 software setup. The emitter saturation current, $J_0e$, is extracted from high-injection generalized photoconduction measurements, as the slope of the generated Auger-corrected lifetime curve.

lifetime wafers. The effective lifetime can, thus, be approximated by

$$\tau_{eff} = -\frac{\Delta n}{\frac{d \Delta n}{dt}},$$ \hspace{1cm} (3.42)

where the steady state generation term from Equation (3.40) has been removed.

A Semilab WT-2000 setup was used to generate spatially resolved carrier lifetime maps during the work with paper II. A schematic drawing of the equipment is shown in Figure 3.14, while a spatially resolved lifetime map can be seen in Figure 3.15.
Figure 3.14: Schematic drawing of a Semilab WC-2000 $\mu$W-PCD system. (Source: Semilab Rt., Budapest, Hungary.)
Figure 3.15: A spatially resolved minority carrier lifetime map of four different samples put back together into the original 4 inch round wafer. The minority carrier lifetime is here shown in a relative scale, where black means long lifetime and red means low lifetime. The low lifetime areas in the left samples result from laser induced damage. The measurements were done with a Semilab WT-2000 equipment. This picture is taken from paper II.
3.6 Thickness measurements

3.6.1 Ellipsometry

We used a variable angle spectroscopic ellipsometer (VASE) from J. A. Wollam Co., Inc. to determine the thickness of single layers of a-Si:H and SiO$_2$ and double-layers thereof. Ellipsometry is commonly used to determine optical properties and thicknesses of thin films. It lends its name from the general, elliptic, polarization state of light. The technique is both versatile, powerful and robust. In principle, the ellipsometer measures the change in polarization of light upon reflection from a sample. The polarization state of the incident light is conveniently decomposed into a transverse (s) and a parallel (p) component relative to the sample surface. The complex reflection ratio, $\rho$, is typically parametrized by $\Psi$ and $\Delta$ as

$$\rho = \frac{r_p}{r_s} = \tan(\Psi)e^{i\Delta},$$

(3.43)

where $r_p$ and $r_s$ are the normalized amplitudes of the p and s components, respectively. $\tan(\Psi)$ is, thus, the amplitude ratio after reflection, while $\Delta$ is the corresponding phase shift. Since the technique only utilizes relative, rather than absolute, values, ellipsometry is insensitive to fluctuations in the light source, and no standard sample or reference beam is needed. However, as the ellipsometry technique is indirect, it is not possible to directly convert the measured values into physical material properties. It requires comparison of simulated data from a model with the measured values. The film thickness is determined by interference between light reflecting from the surface and light traveling through the film(s) and reflecting from the surface between the thin film and the substrate and between thin films in multilayer stacks. The interference can be either constructive or destructive, depending on the relative phase of the rejoining light. In general, the light interference involves both phase and amplitude information, and the phase information obtained from $\Delta$ is sensitive enough to obtain information of films down to sub-monolayer thickness. In our modeling the film thickness was used as a fitting parameter without adjusting the default optical constants for a-Si and SiO$_2$ from Palik[56]. A very good fit between measured and simulated data was obtained for both SiO$_2$ and a-Si. For some samples we also double-checked the
3.7 Contact deposition and resistance measurements

For contact deposition we used DC-magnetron sputtering of aluminum. A plasma is initiated by applying a large voltage across a gap of low-pressure inert gas. Highly energetic ions strike a target containing the material to be deposited. Contrary to the PECVD case, the target, and not the sample, must be placed on the electrode with the minimum ion flux. We have used argon to supply the chamber with an appropriate pressure for deposition, \(3.1 \times 10^{-3}\) mbar. We used a deposition power of 3 kW. The available sputtering system in our lab is an inline system from Leybold Optics.

### 3.7.1 The transfer length method

Solar cell contact resistance, and a technique for its measurement was discussed by Schroder and Meier[92, 93]. We used the so-called transfer length method to extract the contact resistance at the semiconductor-metal interface after laser deposition.
ablation of a-Si:H and a-SiO\textsubscript{x}:H and subsequent sputtering of aluminum. The so-called transfer length, \( L_T \), is a measure of the distance over which most of the current is transferred from the emitter into the contact and vice versa, see Figure 3.17. The relation between \( L_T \) and the specific contact resistance, \( \rho_c \), is given by;

\[
L_T = \sqrt{\rho_c / R_S},
\]

(3.44)

where \( R_S \) is the emitter sheet resistance. The specific contact resistance, \( \rho_c \), arises because of the potential barrier between metal and semiconductor and has units of \( \Omega \cdot \text{cm}^2 \). A more detailed description of the method and the results obtained can be found in paper V.

**Figure 3.17:** Schematic drawing of the current flow in to and out of a contact. \( L_T \) is the so-called transfer length, which is a measure of the distance over which the current is transferred from the emitter (\( p^+ \) in the figure) into the contact pad and vice versa. The Figure is taken from paper V. 2W and 2W’ which are the width of a contact bar and the width between two contact bars, respectively, belong to a more detailed description of the transfer length method and can be found in the paper.
Chapter 4

Results and discussion

In this chapter experimental results that are not discussed in the appended papers are presented. The results are mainly related to process steps that are needed in order to combine the already presented processing steps if they are to be used in the production of a BC-BJ silicon solar cell.

4.1 Diffusion barriers

During the work with this thesis, several different diffusion barriers have been tested with the aim to aid the separation of the doped areas of different polarity on the backside of a BC-BJ silicon solar cell. We have tested the barriers both against phosphorous and boron diffusion. However, the approach we have envisioned for structuring of the backside of the BC-BJ silicon solar cell only requires that the diffusion barrier holds against phosphorous doping as the wafer is assumed uniformly doped with boron on the backside initially to the laser structuring. Early in the process of testing diffusion barriers we sprayed on liquid dopant-containing sources from a custom-made spraying system, see section 3.4. Tested diffusion barriers were PECVD SiO\textsubscript{2} (a-SiO\textsubscript{2}:H) and SiN\textsubscript{x} and a screen-printed TiO\textsubscript{2}-containing paste. An early test of SiN\textsubscript{x} as a diffusion barrier against a phosphorous-containing spray-on source is shown in Figure 4.1. The SiN\textsubscript{x} was deposited through a shadow mask to obtain the desired pattern. The figure shows a spatially resolved sheet resistance map after heavy phosphorus diffusion, and
the interdigitated finger pattern clearly shows the contrast between where the nitride layer was deposited (high sheet resistance) and the areas in between. However, the measurements suffer slightly from the limited resolution of the mapping equipment. The obtained sheet resistance values are thus only to be considered as an indication of the doping level.

Figure 4.1: Spatially resolved sheet resistance map from a Semilab WT-2000. A finger-patterned PECVD SiN\(_x\) layer has been tested as a diffusion barrier against phosphorous doping. The dopant was diluted in solvents sprayed on, dried, and driven-in using a belt furnace.

Both PECVD SiN\(_x\) and screen-printed TiO\(_2\) were found to react chemically with the boron-containing spray-on dopant source, and thus was extremely difficult to
4.1. DIFFUSION BARRIERS

remove after diffusion. This is the reason why we mainly have used SiO$_2$ as both a diffusion barrier and as an etching mask during the work with this thesis. We also continued to use (PECVD) SiO$_2$ after switching to using tube furnaces and gaseous dopant sources to keep the work consistent. However, the experiments in paper II, III, and V could also easily have been done with PECVD SiN$_x$ instead of SiO$_2$. It would just require a reoptimization of the laser parameters. Actually, SiN$_x$ has a better resistance both against KOH etching and phosphorus and boron doping, as it is denser than SiO$_2$.

Figure 4.2 shows the phosphorous concentration in silicon as measured by secondary ion mass spectroscopy (SIMS) after diffusion through SiO$_2$ layers of different thicknesses. With a SiO$_2$ thickness of 450 nm, which is what we mainly have used during the work with this thesis, the phosphorous peak is reduced by 4 orders of magnitude and falls off to the base level (measured on n-type 1-3 Ω-cm wafers) within 40 nm. The small offset of the peak doping from the surface is most likely attributed to redistribution of dopants during a built-in oxidation step in the diffusion process. Fast diffusion of phosphorous in the thermal oxide on the surface will give the profiles seen in Figure 4.2, even if the segregation coefficient of phosphorous is around 10 during oxidation, meaning that the resulting doping concentration in the silicon will be higher than in the oxide[53]. We performed the same diffusion barrier test also for boron doping. The result is shown in Figure 4.3. As can be seen from the figure, the penetration of boron through SiO$_2$ is more significant than that of phosphorous. The segregation coefficient of boron is less than 1, which makes the peak offset from the surface more significant for the unmasked boron profile than for the corresponding phosphorous profile[53]. However, we believe that the absence of an offset for the masked boron-profiles is due to slow oxidation of the surface for low boron concentrations.
CHAPTER 4. RESULTS AND DISCUSSION

Figure 4.2: Phosphorus concentration as a function of depth after diffusion in a tube furnace. Layers of a-SiO$_x$:H with different thicknesses have been tested as diffusion barriers. 0 nm represent an unmasked sample. The results are obtained from secondary ion mass spectroscopy (SIMS).

Figure 4.3: Boron concentration as a function of depth after diffusion in a tube furnace. Layers of a-SiO$_x$:H with different thicknesses have been tested as diffusion barriers. 0 nm represent an unmasked sample. The results are obtained from secondary ion mass spectroscopy (SIMS).
4.2 Contact separation

Different methods exist for contacting of BC-BJ silicon solar cells. The approach we have used resembles somewhat a method developed by Sinton et al. for self-aligned contacts on BC-BJ silicon cells[35]. Their method is based on evaporation of metal on the full backside surface and subsequent chemical contact separation due to different morphologies, and thus etch rates, on the horizontal and vertical planes of the structure, see Figure 4.4. A version of this method is also adapted in the RISE solar cell developed by Engelhart et al. at Institut für Solarenergieforschung, Hameln (ISFH)[39, 94], see section 2.3.

Instead of evaporation we have used in-line sputtering, mainly due to the availability of equipment, see section 3.7. We have used the geometry of the backside structures together with a constantly moving sample carrier to obtain a low metal coverage fraction on the flanks. Subsequently, we used an isotropic aluminum etch to separate the metal contacting the different polarity regions on the backside of the cell by removing the thin aluminum layer on the flanks, see Figure 4.5. The coverage fraction depends solely on the sputtering angle and on the trench depth, see Figure 4.4. A somewhat similar approach was also attempted by Hezel et al., when developing a self-aligned contacting technique by oblique evaporation[95, 96, 97]. Their method was, on the other hand, not originally intended for back-contacted solar cells. A sample with deposited aluminum is shown after etching in Figure 4.6.
**Figure 4.4:** Optical microscope profiling picture of a sample after aluminum deposition by sputtering. The width and depth of the trench are 100 \( \mu \text{m} \) and 12 \( \mu \text{m} \), respectively.
4.2. CONTACT SEPARATION

Figure 4.5: (a) Sketch of the backside of a BC-BJ cell prior to sputtering. The drawing is shown upside-down and is not to scale (b) Same structure after sputtering of aluminum (c) Structure shown after isotropic etching of aluminum. The thin aluminum layer at the flanks is removed, while a layer of reduced thickness remains on the horizontal planes. This process separates the metal contacting the different polarity regions on the backside of a BC-BJ silicon solar cell.
Figure 4.6: SEM picture of a sample after aluminum sputtering and an isotropic etch uniformly removing the aluminum on the sample. The picture shows that the metal is completely removed at the flank, while remaining in the trench.
Chapter 5

Simulating back-junction silicon solar cells

5.1 The Silvaco-ATLAS device simulator

During the work with this thesis, we have used the ATLAS device simulator\cite{98} within the Silvaco simulation framework\cite{99} to model the performance of BC-BJ silicon solar cells. The device simulator enables simulation of electrical, optical, and thermal behavior of semiconductor devices through the finite element method. Our device model has evolved gradually during the whole Ph.D process. The simulations performed during the work with paper I is done with an early version of the device model. Still, the results from the paper give valuable insight into the physical mechanisms important for the performance of a BC-BJ silicon solar cell, and thereby also the design rules that have to be taken into account in the production of such a solar cell. In paper V a more mature version of the ATLAS BC-BJ device model is used together with an external ray tracing program to optimize the semiconductor to metal contact area fraction, and to explore the performance potential of a BC-BJ silicon solar cell design partially based on our experimental results.

When running current-voltage simulations of solar cells, the temperature rise has not been taken into account. That is, we have run the simulations at room temper-
CHAPTER 5. SIMULATING BACK-JUNCTION SILICON SOLAR CELLS

ature (300 K), which is the default in ATLAS. Only during the work with paper III did we include the thermal behavior in the simulations. This was done to study the heat distribution in a silicon sample after a laser pulse. This chapter mainly describes some of the background and the simple basics of the device simulation model, rather than focusing on simulation results.

5.1.1 Defining a unit cell and mesh

Solar cells are relatively large semiconductor devices compared to for instance modern transistors. A commercially available silicon solar module is typically made out of 5 inch wafers, square ( multicrystalline) or semi-square, (monocrystalline) with a thickness of approximately 200 µm. This is, in general, too large to obtain a satisfactory accuracy when simulating the device by the finite element method. We have therefore looked at a the smallest repeating unit within our device, a so-called unit cell, shown in Figure 5.1. A sketch of a complete BC-BJ silicon solar cell is shown in Figure 2.3. The unit cell is sketched without anti-reflective coatings and textured surfaces. This is because the initial ray tracing and calculation of optical losses have been done externally to ATLAS and prior to the electrical simulations, see section 5.1.4. Unit cell simulations give insight into the behavior of the active area of the solar cell. To look at the performance of the total solar cell area one also needs to account for busbars and edge areas. The latter has not been included in the simulations performed during the work with this thesis.

The mesh needs to be sufficiently dense to make sure that the accuracy of the calculations is at a satisfactory level, while at the same time running efficient calculations without unnecessarily long computing times. The mesh density was optimized by repeating calculations with a sequence of finer meshes. A screenshot of a mesh used for simulations in the ATLAS device simulator is shown in Figure 5.2. The mesh has a higher density in the areas where there is high optical intensity and where the gradient of the electric potential is large. A built-in regrid function within ATLAS has been used to allow for a smooth transition of triangle sizes from dense to less dense regions in the mesh.
5.1. THE SILVACO-ATLAS DEVICE SIMULATOR

Figure 5.1: Schematic, simplified drawing of the two-dimensional unit cell used in the simulations of back-contacted back-junction silicon solar cells. The simulations are run in two dimensions. The width in the third dimension is set to 1 $\mu$m by default. BSF and FSF are the back-surface and front-surface fields, respectively. The drawing is not to scale.

5.1.2 Doping

Specification of doping concentrations and profiles can be done in a number of ways in Silvaco-ATLAS. Whole areas, such as a wafer or a unit cell, can be doped to a uniform background level for the wanted base resistivity. Doping profiles of emitter and back- and front surface fields can be given with a gaussian or complimentary error-function profile with specified peak concentration and width or junction depth. This was done in simulations performed during the work with paper I. The profiles can also be read from file to match experimentally measured profiles. In the simulations performed during the work with paper V, we used experimental diffusion profiles as determined with secondary ion mass spectroscopy (SIMS). Examples of SIMS profiles used in the simulation experiments are shown in Figures 4.2 and 4.3 for phosphorus and boron profiles, respectively. Finally, there is also a possibility of doing virtual diffusion experiments within the ATHENA process simulator, which is also a part of the Silvaco simulation framework. These profiles can then be implemented into the device simulations. The latter was tested during the work with paper VI.
Figure 5.2: The mesh of a unit cell used in ATLAS device simulator. The axis units are in µm, and the total mesh consists of approximately 35000 nodes. The density of the grid is significantly higher near the front- and back-surface.

Front-surface field

The advantages of using a front-surface field in BC-BJ silicon solar cells with a pitch compatible with industrial production with low-cost patterning techniques were extensively investigated and discussed by Granek and Hermle[29, 43, 100, 101, 102]. In summary, they conclude that an optimized front surface field

- reduces the influence from the front surface recombination velocity on the solar cell performance,

- improves the UV-stability of the cells compared to BC-BJ silicon solar cells without a FSF,
• improves the low illumination performance compared to cells without a FSF, and

• reduces the lateral series resistance of majority carriers.

They found that a deep gaussian profile gave the best results. In accordance with the results of Granek and Hermle, we have implemented a FSF with a Gaussian profile, a peak doping concentration of $3.8 \times 10^{18}$ cm$^{-3}$, and a junction depth of 1.44 µm. The sheet resistance of the a profile is 148 Ω/□.

**Base resistivity**

McIntosh et al. investigated the influence of base resistivity on the performance of BC-BJ silicon solar cells[36]. They found that the optimum base resistivity of a BC-BJ cell processed with low-cost patterning techniques should be close to 2 Ω·cm for a wafer in the thickness range 160-280 µm. The efficiency of the cells did not drop significantly up to a base resistivity of 10 Ω·cm. The base doping is mostly a trade off between high short-circuit current density due to low recombination of minority carriers at high wafer resistivity, and high open-circuit voltage and low base series resistance at low wafer resistivity. We have therefore run most of our simulations with a wafer base doping of $2.4 \times 10^{15}$ cm$^{-3}$, corresponding to a base resistivity of 2 Ω·cm for n-type wafers.

**5.1.3 Contacts**

Since the contact interface between metal and semiconductor is highly doped both for the emitter and the BSF ($\approx 10^{20}$ cm$^{-3}$), we have defined our contacts as ohmic electrodes rather than a specific material (aluminum in our case). This is done to avoid that the contacts are being treated as Schottky barriers, and thus unnecessarily increase the computing times. The current-flow in a BC-BJ silicon solar cell, and thus into the contacts, is substantially lateral. Therefore, a distributed (specific) contact resistance has been specified. The specific contact resistance is defined as the slope of the $J$-$V$ curve at $V = 0$:

$$\rho_c = \left\{ \frac{\partial V}{\partial J} \right\} |_{V=0}.$$  \hspace{1cm} (5.1)
CHAPTER 5. SIMULATING BACK-JUNCTION SILICON SOLAR CELLS

In ATLAS this is calculated as

$$R_i = \frac{\rho_c}{w \cdot d_i}.$$  \hspace{1cm} (5.2)

$R_i$ is the resistance associated with node $i$ at the surface, $\rho_c$ is the user specified distributed contact resistance, $d_i$ is the width of the surface segment associated with node $i$, and $w$ is the width of the unit cell in the third dimension. The specific contact resistance used in the device simulation model during the work with paper V was set to 0.8 mΩ·cm², based on experimental results from the same paper. The contacting was done by sputtering of aluminum through laser ablated holes in a double-layer of a-Si:H and a-SiOₓ:H, see paper III, IV, and V.

5.1.4 Optical properties

Ray Tracing

The evaluation of optical loss mechanisms was performed externally to ATLAS. This was done due to limited possibilities of performing ray tracing with the full solar spectrum on structured surfaces in ATLAS. The ray tracing simulator TracePro[103] was used to calculate the optical losses due to reflection, parasitic absorption, escaped light, and free carrier absorption (FCA). The front surface was assumed perfectly covered with upright pyramids with a base width of 4 µm and an angle of 54.7° of the side walls to the surface orthogonal. The pyramids, which are typically fabricated by etching of the silicon surface in a weak alkaline solution, are introduced to lower the overall surface reflections. This is the industry standard for monocrystalline silicon solar cells. The pyramids were assumed covered with an anti-reflection coating of 10 nm SiO₂ close to the surface and 60 nm SiNₓ on top of the oxide. The thickness of the silicon wafer was set to 200 µm. On the backside, the original doping profiles were replaced by 300 nm deep uniform box profiles of phosphorus and boron. This was done to simplify the simulations while at the same time ensuring that the overall integrated FCA in the uniform profiles coincides with that of the actual doping profiles used in the device simulations. The unit cell used for the device simulation when also including the optical loss calculations is shown in Figure 5.3.
5.1. THE SILVACO-ATLAS DEVICE SIMULATOR

We chose to cut off the optical simulations at 1100 nm due to very long absorption lengths, and thus computing times, above this wavelength. However, the intensity of the AM 1.5 solar spectrum is relatively low between 1100-1150 nm, and the error introduced by this cut-off is expected to underestimate the short-circuit current density with around 0.2 mA/cm$^2$. At the same time, ray tracing has been reported to overestimate the anti-reflection effect of the pyramids[104]. The overestimation originates from the fact that light is not able to resolve the geometrical sharp corners of the textures, thereby allowing some direct transmission or reflection, not found in the ray tracing[104]. The main results from the ray tracing are shown in Figure 5.4. Primary reflection is the limiting optical loss mechanism at short wavelengths, while escaped light is dominating at long wavelengths.

Figure 5.3: Schematic drawing of the unit cell used for the two-dimensional device simulations when also including ray tracing. The optical losses were calculated externally to ATLAS and prior to device simulation with the program Trace-Pro. FSF and BSF are the front- and back-surface fields, respectively. The figure is simplified and not to scale.

Free carrier absorption

Free carrier absorption (FCA) is an intrinsic absorption process in silicon that involves transitions between free electrons or holes. It does not create an additional electron-hole pair, and can be a significant optical loss mechanism in some solar cell designs. A expression of FCA commonly used for solar cells was
parametrized by Green [105];

$$\alpha_{FC} = 2.6 \times 10^{-18} \times n \times \lambda^3 + 2.7 \times 10^{-18} \times p \times \lambda^2,$$

(5.3)

where $n$ and $p$ are the electron and hole concentrations (in cm$^{-3}$), respectively, while $\lambda$ is the wavelength in $\mu$m. FCA as function of wavelength is shown in Figure 5.5.

![Figure 5.4: The fraction of light absorbed as a function of wavelength as calculated with ray tracing. The major optical loss mechanisms are also indicated in the figure.](image)

**The modified AM 1.5 solar spectrum**

The calculated absorbed light was weighted against the standard AM 1.5 solar spectrum[24], see Figure 5.6. This was done by interpolating the absorbed light data points to obtain the same density of data points as in the AM 1.5 solar spectrum in the same wavelength range. Each data point in the original spectrum was then multiplied with the corresponding fraction of light absorbed at the same wavelength to obtain the modified spectrum. Finally, the modified spectrum was used as an input to the ATLAS device simulator in Silvaco.
**Figure 5.5:** Free carrier absorption as function of wavelength for different doping concentrations according to Equation (5.3). The solid lines show FCA in n-doped silicon, while the dashed lines show the corresponding FCA in p-doped silicon. The absorption coefficient for intrinsic silicon is also shown for comparison.

### 5.1.5 Interfaces

**Surface reflection and transmission**

Since the incident solar spectrum already has been modified to account for primary reflection losses, the front surface is assumed non-reflective for incident illumination. Internally, however, the surfaces are now assumed fully reflective since also the escaped light has been accounted for in the ray tracing program. In other words, of the incoming modified spectrum nothing is reflected, transmitted, or lost optically in any way.
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Figure 5.6: The standard AM 1.5 solar reference spectrum shown together with the modified solar spectrum used as input to the ATLAS device simulator. The modified spectrum is corrected for the optical losses such as surface reflection, escaped light, metal parasitic absorption, and free carrier absorption in the solar cell. The short wavelength losses are mainly due to primary reflection while the reduced irradiance at long wavelengths is caused by escaped light, free carrier absorption, and backside metal absorption.

Surface recombination

The front-surface recombination velocity has been calculated according to Cuevas et al.[106] assuming a thermally oxidized surface;

\[
S = S_0 \quad \text{for } N_D < N_{\text{ref}}
\]
\[
S = S_0 (N_D / N_{\text{ref}}) \quad \text{for } N_D \geq N_{\text{ref}},
\]

where \( S_0 = 70 \text{ cm/s} \) and \( N_{\text{ref}} = 7 \times 10^{17} \text{ cm}^{-3} \), while \( N_D \) is the phosphorus doping concentration. On the backside, we have assumed the surface covered by a double-layer of passivating a-Si:H and a-SiO\(_x\):H, see papers III and V. The surface recombination velocity has been calculated according to Aberle et al. as[78];

\[
S_{\text{eff}} = \frac{J_{\text{sc}} N_A}{n_i^2},
\]
where $J_{oe}$ is the emitter saturation current density, and input from experimental results obtained in paper III. The metal-semiconductor interface, i.e. the contact points, have been modeled with a surface recombination velocity of $3 \times 10^6$ cm/s[107].

### 5.1.6 Physical models

#### Bulk recombination

We used the default ATLAS model for Auger recombination which has been implemented according to Dziewior et al. as[108]

$$U_{\text{Aug}} = A_p (n^2 p - n_0^2 p_0) + A_n (n^2 p - n_0^2 p_0)$$  \hspace{1cm} (5.6)

where $n$ and $p$ and $n_0$ and $p_0$ are the concentrations and equilibrium concentrations of electrons and holes, respectively. The Auger coefficients, $A_p$ and $A_n$, are set to $9.9 \times 10^{-32}$ cm$^6$/s and $2.8 \times 10^{-31}$ cm$^6$/s, respectively.

In addition, we used a concentration dependent Shockley-Read-Hall recombination model which is also implemented to the ATLAS simulation software[109, 110, 111];

$$U_{\text{SRH}} = \frac{np - n_i^2}{\tau_{n,\text{SRH}}(p + p_t) + \tau_{p,\text{SRH}}(n + n_t)}.$$  \hspace{1cm} (5.7)

$n_t$ and $p_t$ are the densities of electron and hole traps, respectively. $\tau_{n,\text{SRH}}$ and $\tau_{p,\text{SRH}}$ are the Shockley-Read-Hall lifetimes of electrons and holes, respectively, depending on the local doping concentration.

#### Carrier mobility

The carrier mobility has been specified by using a look-up table model to relate the low-field mobility at 300K to the impurity concentration[98]. The table contains empirical data for the carrier mobility of Silicon (only) at different doping concentrations.
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Apparent bandgap narrowing

In the presence of heavy doping, experimental results have shown that the \( p \)\( n \) product in silicon becomes doping dependent\[112\]. This is in complete analogy with the law of mass action in ideal gasses. The relation \( n_i^2 = n_0 p_0 \) is not valid if the charge carries interact strongly with each other. Several different formulae for the decreased bandgap in silicon devices due to heavy doping have been suggested\[113, 114, 115, 116\]. However, an parametrization of Cuevas et al. has found widespread use in the solar cell community\[107\];

\[
\Delta E_G = A \ln\left(\frac{N}{N_{\text{ref}}}\right).
\]

\( \Delta E_G \) is here the change in bandgap due to the doping concentration, \( N \). \( A \) is a constant

\[
A = 0 \quad \text{if} \quad N < N_{\text{ref}}
\]

\[
A = 14 \text{ meV} \quad \text{if} \quad N \geq N_{\text{ref}},
\]

and \( N_{\text{ref}} = 1.4 \times 10^{17} \text{ cm}^{-3} \). The change in bandgap influences the carrier statistics through a change in the intrinsic carrier concentration;

\[
n_i = n_{i0} \exp\left(\frac{\Delta E_G}{2kT}\right)
\]

It can be shown that this leads to a change in the effective dopant concentration as \[117\];

\[
N_{\text{eff}} = N^{1-A/kT} N_{\text{ref}}^{(A/kT)}.
\]

The most commonly used bandgap narrowing models referenced above are all empirical models deduced from transport measurements of highly doped silicon. Physical effects like carrier-carrier and carrier-impurity interactions are not separated from each other\[118\]. The model of Cuevas, which is the one we have used, includes several heavy doping effects. The combined total heavy doping effect is called the apparent bandgap narrowing. The model has therefore implicitly built in the effect of Fermi-Dirac statistics which also becomes significant at high doping densities. Fermi-Dirac statistics and apparent bandgap narrowing models can
therefore not be used simultaneously in a consistent way. In the one-dimensional solar simulator program PC1D the BGN model of Cuevas is used but requires simple Boltzmann statistics to produce accurate results. A more consistent bandgap narrowing model derived from quantum mechanical principles was developed by Schenk[119] and implemented in the Dessis device simulator for simulation of solar cells by Schumacher et al.[118]. In our case this is beyond the scope, and Cuevas’ model suffices for now.

It is also worth noting that in paper I, which served as a preliminary investigation both of BC-BJ silicon solar cells and also the ATLAS device simulator itself, we used Slotboom’s BGN model together with Fermi-Dirac statistics. As can be seen from Figure 5.7 there is a rather large discrepancy between the different models for BGN. Slotboom’s model, which is the default in ATLAS, implies a more significant bandgap decrease than the other models. Even del-Alamo’s model, which implies a more modest narrowing effect than the model of Slotboom et al., has been shown to overestimate the decrease in the bandgap for very heavy doping[118]. The use of Fermi-Dirac statics models will reduce carrier-carrier interactions, and thereby recombination. It is thus opposing the effect of bandgap narrowing. Even if it is inconsistent to use fermion carrier statistics together with empirical mod-

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**Figure 5.7:** Apparent bandgap narrowing as a function of doping concentration according to different commonly cited references.
els for bandgap narrowing, the effect of using Fermi-Dirac statistics was found to more or less cancel out the overestimated narrowing from Slotboom’s empirical apparent bandgap narrowing, see Figure 5.7.
Figure 5.8: (a) Current-density \((J, \text{ absolute value})\)-voltage \((V)\) curves resulting from the activation of different combinations of bandgap narrowing models with different carrier statistics models. The solid lines show \(J-V\) curves with the use of Fermi-Dirac carrier statistics, while the dashed lines show simulations performed with simple Boltzmann carrier statistics. The blue lines indicate the use of Slotboom’s bandgap narrowing model, while the green lines where simulated with the use of the bandgap narrowing model of Cuevas et al. (b) The local ideality factor, \(m\), as function of applied voltage. The same combinations of line-color and style as described above also applies for the \(m-V\) curve. For a brief introduction to \(m-V\)-curves, see Section 5.2.
CHAPTER 5. SIMULATING BACK-JUNCTION SILICON SOLAR CELLS

5.2 \( m-V \) curves

A powerful method to investigate the characteristics of a solar cell is to study the local ideality factor, \( m \), as a function the applied voltage, \( V \). \( m \) is related to the slope of the \( J-V \) curve, and can often reveal information that is not directly available from that curve. We let \( m \) be a free parameter so that

\[
m = \frac{1}{V_{th} d(V_{th} \ln J_S)}. \tag{5.12}
\]

\( V_{th} \) is the thermal voltage (see section 2.1), and \( J_S \) is the shifted current density, i.e. \( J \) minus the photogenerated current, \( J_{SC} \);

\[
J_S(V) = J_0 \left[ \exp((V + JR_S)/V_{th}) - 1 \right] + \frac{V + JR_S}{R_{SH}}. \tag{5.13}
\]

We have assumed that the light-generated current density is equal to the short-circuit current density, which is usually a good approximation. It is thus only distinguished from the dark current density by the sign before the voltage drop induced by the series resistance, \( (JR_S) \);

\[
J_{dark}(V) = J_0 \left[ \exp((V - JR_S)/V_{th}) - 1 \right] + \frac{V - JR_S}{R_{SH}}. \tag{5.14}
\]

This is a fundamental difference between \( p-n \) diodes in the dark and under illumination. Under illumination the total voltage is the sum of the applied voltage over the terminals and the voltage drop over the series resistance of the device. In the dark it is the applied voltage that drives the current, and the total voltage is the difference between the applied voltage and the voltage drop over the cell. This can be seen as a shift in the \( m-V \) curve between the shifted and dark \( J-V \) characteristics, see Figure 5.9. In practical devices, \( R_S \) and \( J_0 \) will in general be different in the dark and under illumination. The free parameter, \( m \), must not be confused with the ideality factor that is associated with single recombination mechanisms. The resulting \( m(V) \) comes from a combination of different recombination mechanism with different exponential behavior.
5.3 Tunneling

In a heavily doped p-n diode, charge carriers can tunnel through the junction barrier despite their insufficient energy. Tunneling through the bandgap can take place directly between bands on opposite sides of the barrier, so-called band-to-band tunneling (BBT)[120, 121], or via traps in the bandgap, so-called trap-assisted tunneling (TAT)[122], see Figure 5.10. For conventional front-contacted silicon solar cells this is not of any importance, as only one side of the p-n junction is heavily doped. In multi-junction solar cells controlled interband Esaki tunneling is used to let the current pass freely between the subcells[123, 124].

In solar cell designs where highly doped regions of opposite polarity are present on the same side, quantum mechanical tunneling of charge carrier can be a dominating effect. In combination with back-contacted silicon solar cells, tunneling and the effect of tunneling on the solar cell performance is not well described. Guo et al. discussed how tunneling could have an detrimental effect on the fill factor of double-side buried contact (DSBC) solar cells, but it was concluded that...
tunneling was not a likely cause of the low fill factor[125]. Both the Sunpower A-300 silicon solar cell and the RISE silicon solar cell of ISFH have highly doped regions of opposite polarity in direct contact with each other as a consequence of the design and production[38, 39], see section 2.3. In order to get a better understanding of how the tunneling effect can influence the performance of BC-BJ silicon solar cells, we have performed simulations of BC-BJ silicon solar cells including lateral tunneling through the highly doped p-and n regions on the back of the cell, see Figure 5.11.

In order to be able to simulate the tunneling effect, a very fine rectangular mesh was superimposed on the original grid. The lateral distance between the grid points was set to 30 picometers in the tunneling direction.

In Figure 5.12 the $J-V$ curves of a small illuminated tunneling junction is shown when there is a doping density of $5 \times 10^{19} \text{ cm}^{-3}$ on both sides of an abrupt junction. The effect of the different doping mechanisms that dominate in different voltage ranges are clearly seen in the figure. The characteristic negative different-

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**Figure 5.10:** Schematic drawing of tunneling mechanisms in a highly doped p-n junction.
Figure 5.11: Schematic drawing of a BC-BJ silicon solar cell, including the tun-
neling junction. FSF and BSF are the front and back-surface fields, respectively.

tial resistance region can be seen as a bump at low forward applied voltage, and is
caused by BBT. Trap-assisted tunneling is the dominant tunneling mechanism in
the forward applied voltage region around the maximum powerpoint, indicating
that it will have a negative influence on the performance of a silicon solar cell. The
effect of reducing the doping concentration to $1 \times 10^{19} \text{ cm}^{-3}$ can be seen in Fig-
ure 5.13. The negative resistance region is no longer visible in the curve, but the
current density in the forward applied voltage region is still dominated by TAT.
A more thorough discussion of quantum mechanical tunneling in back-junction
silicon solar cells, can be found in paper VI.

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Figure 5.12: Current-density (absolute value)—voltage curves of an illuminated tunneling junction. The doping concentration was $5 \times 10^{19}$ cm$^{-3}$ on both sides of an abrupt junction. The different curves are result from simulations of the same p-n junction but with different tunneling mechanisms enabled. BBT and TAT are abbreviations for band-to-band and trap-assisted tunneling, respectively.

Figure 5.13: Current-density (absolute value)—voltage curves of an illuminated tunneling junction. The doping concentration was $1 \times 10^{19}$ cm$^{-3}$ on both sides of an abrupt junction. The different curves are result from simulations of the same p-n junction but with different tunneling mechanisms enabled. BBT and TAT are abbreviations for band-to-band and trap-assisted tunneling, respectively.
Chapter 6

Conclusions

In this thesis I have studied several important issues regarding back-contacted back-junction (BC-BJ) silicon solar cells. The work has been done both through experiments and simulations. The results are mainly presented in the six appended papers, while the background for the cell concept, experiments, and simulations are covered in the previous chapters of the thesis. In addition, two patents have been filed based on findings presented herein. The main conclusions are summarized and presented below.

In the experimental work I have focused on the use of laser as a processing tool. A diode pumped, frequency doubled, Nd:YVO$_4$ green (532 nm) laser with pulse duration in the nanosecond regime was used in all the experiments. This laser is industrially well proven and relatively cheap. I showed how large areas of silicon could be ablated (removed) from a monocrystalline silicon wafer with laser and subsequent etching in an aqueous KOH solution without degrading the minority carrier lifetime in the processed areas. The ablation was done with the aim of structuring a BC-BJ silicon solar cell, and high minority carrier lifetimes are necessary in order to obtain high conversion efficiencies in this solar cell design. A laser pulse fluence (energy density) of 2.0 J/cm$^2$ at a repetition frequency of 20 kHz with a spot to spot overlap of 75 % were found to give the best results with subsequent etching in an aqueous KOH solution with a concentration of 55 % (weight) at 80 ºC. I also demonstrated how the same laser could be used to make
holes in a double passivating layer of amorphous silicon (a-Si) and silicon oxide (SiO$_2$) without damaging the underlying crystalline silicon wafer. The holes were opened with the aim of making local contacts on the back of a BC-BJ silicon solar cell. I explained how the high optical absorption and low thermal conductivity of a-Si were the key parameters to enable ablation of the covering SiO$_2$ with laser fluences five times lower than that needed to ablate crystalline silicon (c-Si) with the same laser. I also showed how the laser fluence threshold for SiO$_2$ ablation decreases with increasing thickness of the a-Si layer up to a thickness corresponding to the thermal diffusion length of a-Si. Furthermore, I looked in more detail on the interaction between laser and thin film a-Si on c-Si, and developed a thermodynamic model that was in good agreement with the previous experimental results. The model was able to predict the laser fluence threshold for SiO$_2$ ablation as a function of the a-Si thickness. The modeling also gave insight into the importance of the thermal resistance at the interface between a-Si and c-Si for the successful use of the laser in the damage-free ablation process. A thermal interface resistance of 0.0054 cm$^2$·K/W gave the best match between the model and the experimental results. This resistance corresponds to an additional layer of native oxide between the two materials.

I investigated the quality of the laser ablated holes in the passivating double-layer of a-Si/SiO$_2$ by deposition of aluminum and use of the transfer length method to extract the specific contact resistance in the laser ablated holes. A specific contact resistance as low as 0.8 mΩ·cm$^2$ was measured for an a-Si layer with a thickness of 180 nm and a laser fluence of 0.59 J/cm$^2$. Despite our relatively standard laser, these results are better than what has been obtained by most state-of-the-art laser technologies using lasers with pulse durations in the pico- and femtosecond range. In order to investigate how the optimized specific contact resistance influences the performance of a full BC-BJ cell, I simulated the active area performance of a cell given the latest experimental results. The simulation model also included full optical loss analysis and so gave a good indication of the efficiency potential of the cell structure I have been working with. Given realistic values for material quality, physical parameters, and the size of the backside structures; I ended up with a simulated active area efficiency of 20.4 % on a 200 μm thick 2 Ω·cm n-type silicon wafer. The diffusion profiles used in the simulations were based on experimental,
and not fully optimized, results. There is also room for further improvement of the short-circuit current density by introducing slightly more advanced, but potentially costly, anti-reflection coating schemes than what was assumed in the ray tracing. Finally, tunneling mechanisms in the adjacent highly doped regions on the backside of a BC-BJ silicon solar cell was investigated through both device, process, and circuit simulations. I showed how so-called trap-assisted tunneling can have a deleterious effect on the efficiency of a silicon solar cell through a reduced shunt-resistance, and thus fill-factor. Band-to-band tunneling, on the other hand, does not influence the performance of a solar cell near the preferred operating point on the current-voltage curve, and so does not effect the efficiency. Band-to-band tunneling does, however, lead to current breakdowns at low to moderate reverse voltages. This implies that potential hotspots can be avoided under reverse voltage, as the heat dissipation will be distributed throughout the whole junction region. This cell type therefore has the potential of operating in a solar module without the need for bypass diodes.

Still, much work remains to be able to finalize a full version of a BC-BJ silicon solar cell. Processing steps related to the front-side have not been studied (experimentally). Texturization, anti-reflection coatings, passivation, and front-surface fields are all important parts of a successfully working back-junction cell. The metallization scheme used herein has been found suitable for small lab-scale solar cells, however, for large-area cells alternatives or a revised version of the sputtered aluminum need to be investigated.

Finally, it is also important to point out that most of the results obtained herein might separately be as useful for other solar cell designs as for BC-BJ silicon solar cells. Especially the idea of using a-Si as a combined passivation and laser irradiation buffer layer should be attractive for local contacting in both high-efficiency front-contacted cells as well as in other back-contacted silicon solar cell designs.
Appendix

Erratum

In paper II we investigated ablation of large areas of silicon from a silicon wafer. The work was done prior to proper measurements of laser beam characteristics. We therefore assumed the $\frac{1}{e^2}$ diameter of the laser spot to be 20 $\mu$m in focus, based on information from the laser system manufacturer. Later, by measuring the laser spot diameter by a BC106-VIS CCD camera beam profiler from Thorlabs, we found the actual $\frac{1}{e^2}$ diameter to be 40 $\mu$m in focus, and thus the information from the manufacturer erroneous. This implies that the optimized laser parameters obtained in paper II of 50 % spot to spot overlap of a 20 $\mu$m wide laser spot, in reality was 75 % spot to spot overlap of a 40 $\mu$m wide laser spot, see Figure 6.1. This does not influence the final results in any way. The correct laser spot characteristic are used in the papers following paper II.
Figure 6.1: **Left:** Assumed 50% spot to spot overlap for a laser spot with a diameter of 20 $\mu$m. **Right:** The actual 75% spot to spot overlap for a laser spot with a diameter of 40 $\mu$m.
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Appended papers
Paper I
NUMERICAL OPTIMIZATION OF A BACK-JUNCTION SOLAR CELL

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ABSTRACT: This paper presents work on the optimization of a back-junction solar cell using the Silvaco-Atlas simulation framework. The unit cell width (the distance between the midpoints of two neighboring electrodes), the wafer thickness, the back-side emitter coverage fraction, and the effective minority carrier lifetime are varied to find an optimized set of parameters for an interdigitated p- and n-finger rear-side design. Other variables are held constant. We show that the back-side emitter coverage fraction in principle should be as large as possible to increase the short-circuit current, but due to practical limitations we assume that at least 20% of the unit cell area should be left for the back-surface field/base contact area. The optimal wafer thickness as well as the optimal unit cell width decreases with decreasing effective minority carrier lifetime. At high effective lifetimes the open-circuit voltage is shown to decrease with decreasing unit cell width and thus reducing the limiting efficiency for small unit cells. For shorter minority carrier lifetimes, however, the decrease in open circuit voltage is much smaller and the corresponding increase in short-circuit current density results in increased efficiency when reducing the unit cell width with this design. A large unit cell (at the order of 500 µm) is therefore preferable if high quality wafers with excellent surface passivation is available. On lower quality materials, with lower minority carrier lifetimes, a reduced unit cell size will increase the solar cell efficiency.

Keywords: Simulation, Back-Junction, Silvaco

1 INTRODUCTION

Our research institute is currently developing a high-efficiency back-contacted solar cell. The research is carried out both through experiments and simulations. The Silvaco simulation framework [1] is used to do advanced and accurate simulations of both process and device. At the same time we have two parallel experimental processes running. One clean room and one industrial like process. This paper describes the numerical simulation and optimization of a back-junction solar cell with interdigitated p- and n-fingers.

The Silvaco-Atlas device simulator has been shown to be a powerful tool for simulating solar cells [2-4]. Under a given solar spectrum the simulator will numerically reproduce the behavior of a solar cell by solving the Poisson equation, the carrier continuity equation, the drift-diffusion transport model, and the energy balance at each individual node in the mesh. In our simulations we vary several different key parameters, one at the time, to find an optimized parameter set. These results are to be used in upcoming experimental work. The variables studied here are the wafer thickness, the unit cell width (UCW, the distance between the midpoints of two neighboring electrodes), the emitter coverage fraction (ECF) on the rear-side, and finally the effective minority carrier lifetime (MCL). The efficiency, the short-circuit current density (Jsc), the open-circuit voltage (Voc), and the fill-factor are extracted from the IV-curve for each parameter set.

2 BACK-JUNCTION SOLAR CELLS

Back-contacted solar cells have served as an alternative to conventional solar cells for more than 30 years [5], and many different cells with quite different designs, such as metallization-wrap-through, emitter-wrap-through, and back-junction cells, have been developed. Back-junction solar cells have both the emitter and the complete metallization on the rear side. This eliminates the shading present in conventional solar cells, and the sunward and back-side can therefore be independently optimized for optical and electrical performance, respectively. Moreover, having both contacts on the rear-side significantly simplifies module assembly, and it allows for an increased packing density of the cells. This type of cell design is usually believed to require high quality monocrystalline silicon wafers with long MCLs and excellent front surface passivation, since the carriers, created predominantly near the front surface, have to diffuse through the whole wafer. Our results show that lower quality materials also can give good results as long as the wafers are thin enough. An extensive review on back-contacted solar cells can be found in ref. [6].

3 METHODOLOGY

Since both contacts are placed on the rear side of the cell, there is a significant amount of lateral current, and a comprehensive analysis of such a structure needs to take into account two or even three dimensional effects, depending on the emitter geometry. In this work we look at interdigitated emitter and base fingers and thus content ourselves with two dimensional simulations.

3.1 Defining a structure

We have defined the unit cell in our 2 D model to be the distance between the center of two neighboring base and emitter contacts. In experiments, the minimum distance between fingers will be limited by the possible alignment resolution of the process equipment. A typical clean room process with access to photo-lithography will have the possibility of making much finer structures than an industrial like process. The width of the unit cell is varied between 100 µm and 500 µm in this work, and
thus covers the possible alignment resolution of a variety of experimental processes.

The specification of a mesh always involves a trade off between accuracy and numerical efficiency. Compared to other semiconductor devices that typically are studied numerically in this kind of simulation software, the unit cells we work with here are rather large, which implies that a fine but at the same time efficient mesh is of crucial importance. An equally dense mesh over the whole unit cell would result in prohibitively long computing times. In Atlas there is a possibility of regridding the mesh with respect to various variables. The regrid command will automatically refine the mesh in a certain region if a specified solution variable exceeds some value, or when the change of that variable across the area between nodes exceeds some predefined value. By repeating a calculation using a sequence of finer meshes the error associated with the structure has been investigated systematically. The front of the cell is assumed non-reflecting and perfectly passivated in these simulations.

3.2 Physical models

There are numerous physical-numerical models built into the Silvaco-Atlas framework. Only a few are activated by default, and other models have to be carefully chosen. Concentration dependent Shockley-Read-Hall recombination, concentration dependent mobility, Auger recombination, band-gap-narrowing, Fermi-Dirac statistics, and field-dependent mobility are the models we have included in these simulations. Figure 1 shows the effect of changing from Boltzmann- to Fermi-Dirac statistics on a solar cell. The change in efficiency is mainly due the reduced recombination as a consequence of the Pauli Exclusion Principle.

3.3 Parameter settings

Table I shows the parameters used in the simulations of our back-junction solar cell.

Table I: Parameters for the simulated structure

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>Base</td>
<td></td>
</tr>
<tr>
<td>Type</td>
<td>n</td>
</tr>
<tr>
<td>Thickness</td>
<td>50-300 μm</td>
</tr>
<tr>
<td>Doping conc.</td>
<td>3.5x10^{15} cm^{-3}</td>
</tr>
<tr>
<td>Min. carrier lifetime</td>
<td>62 μs - 2ms</td>
</tr>
<tr>
<td>BSF</td>
<td></td>
</tr>
<tr>
<td>Type</td>
<td>p</td>
</tr>
<tr>
<td>Doping conc.</td>
<td>5x10^{19} cm^{-3}</td>
</tr>
<tr>
<td>Depth</td>
<td>300 nm</td>
</tr>
<tr>
<td>Profile</td>
<td>Gaussian</td>
</tr>
<tr>
<td>Electrode</td>
<td></td>
</tr>
<tr>
<td>Width</td>
<td>30 μm</td>
</tr>
<tr>
<td>Position</td>
<td>Center of p- or n-finger</td>
</tr>
<tr>
<td>Unit Cell</td>
<td></td>
</tr>
<tr>
<td>Width</td>
<td>100-500 μm</td>
</tr>
<tr>
<td>Position</td>
<td>Center to center p-to-n-electrode</td>
</tr>
<tr>
<td>Surface Recombination</td>
<td></td>
</tr>
<tr>
<td>Front</td>
<td>0 cm/s</td>
</tr>
<tr>
<td>Back</td>
<td>0 cm/s</td>
</tr>
</tbody>
</table>

4 RESULTS AND DISCUSSION

If not stated otherwise, we assume that the effective MCL is 2 ms, the emitter and BSF doping concentrations are 5 x 10^{19} cm^{-3} (p- and n-type, respectively), and that the wafer thickness and the UCW both are 300 μm.

4.1 Wafer thickness and bulk lifetime

An optimized wafer thickness is especially important in this kind of cell design since the carriers generated at the front have to diffuse through the whole cell. Figure 2 shows J_{SC} as a function of wafer thickness for different effective minority carrier lifetimes. The optimum wafer thickness depends strongly on the bulk material and on the quality of the front- and back surface passivation. With an effective MCL of 2 ms (corresponding to a diffusion length of 1560 μm) the limiting J_{SC} is 39.33 mA/cm² for a 300 μm thick wafer. At 62 μs effective MCL (corresponding to a diffusion length of 270 μm) J_{SC} drops to 24.11 mA/cm² for the same wafer thickness. If we are able to obtain a 2 ms effective lifetime a maximum J_{SC} is found at a wafer thickness close to 300 μm. Shorter effective MCLs give optimized currents at thinner wafers. For instance, at an effective lifetime of 125 μs the maximum short circuit current density is obtained with wafers with a thickness close to 100 μm. In the extreme case where L>>T (the diffusion length of the minority carriers is much larger than the wafer thickness) J_{SC} will approach a limiting value proportional to 1-e^{-1/T} (T is really the thickness of the wafer minus the depth of the
highly doped rear side and the depletion zones). In the other extreme case; L<<T (the diffusion length is much shorter than the wafer thickness), the short circuit current will fall off proportional to $e^{-T}$ [7].

![Figure 2](image)

**Figure 2:** The figure shows the short circuit current density as a function of the wafer thickness for different effective minority carrier lifetimes for a 300 μm wide unit cell with 80% emitter coverage fraction.

$V_{OC}$ is also a strong function of lifetime, shown in Figure 3. For high MCLs $V_{OC}$ is slightly reduced when increasing the wafer thickness while for shorter effective lifetimes the voltage drop is more significant. The limiting open circuit voltage is given as $V_{OC} = (kT/q) \ln (pN_d/n_i^2)$, where $N_d$ is the base doping concentration, $n_i$ is the intrinsic carrier concentration and $p$ is the injected hole concentration. The drop in $V_{OC}$ results from the recombination of minority carriers (holes) which is much more pronounced for shorter diffusion lengths and thicker wafers. At 62 μs bulk lifetime the open-circuit voltage decreases from 0.640 V to 0.589 V when increasing the wafer thickness from 50 μm to 300 μm. For a 2 ms lifetime sample the corresponding reduction in $V_{OC}$ is only 2 mV; from 0.665 V to 0.663 V.

![Figure 3](image)

**Figure 3:** The figure shows the open-circuit voltage as a function of the wafer thickness for different minority carrier lifetimes for a 300 μm wide unit cell with 80% emitter coverage fraction.

The fill-factor is also affected by both lifetime and wafer thickness. An increasing $V_{OC}$ tends to increase the Fill-factor while an increasing short-circuit current has the opposite effect. The latter is due to an increasing series resistance. As a result, the fill-factor increases with increasing wafer thickness for the 62 μs effective MCL due to the significant drop in $J_{SC}$, but for all other simulations (i.e. lifetime larger than 62 μs) it decreases with increasing wafer thickness. The influence of $V_{OC}$ and $J_{SC}$ on the fill-factor results in a minimum for a specific MCL for a fixed unit cell width. This effect was also reported in ref. [8].

### 4.2 Emitter coverage fraction and unit cell width

We have also run simulations to find an optimal rear-side ECF. The rest of the back is assumed covered by a back-surface field (BSF). In principle the ECF should be as large as possible, but in practical devices this is limited by the minimum alignment resolution of the process equipment and the width of the contact point on the n+ BSF. For an industrial like process a realistic equipment alignment resolution will be close 200 μm, which implies that a BSF coverage fraction of less than 20% will be unrealistic to align for unit cells smaller than 500 μm. In a clean room process with access to photolithography, however, we do not have this limitation and a UCW of 100 μm and below is easily achieved. Figure 4 shows the limiting $J_{SC}$ as a function of ECF for different UCWs. The current density increases with increasing coverage fraction for all simulations. This is easily explained by the reduced recombination due to the reduced average lateral distance that a minority carrier has to travel in order to reach the base/emitter junction. The effect of increasing the emitter coverage fraction on the short-circuit current is more pronounced for larger unit cells and shorter effective MCL.

![Figure 4](image)

**Figure 4:** The figure shows the limiting short-circuit current density as a function of the p+ emitter fraction for different unit cell widths.

In Figure 5 the open circuit voltage is shown as a function of the unit cell width for different ECFs. As for $J_{SC}$, $V_{OC}$ increases with increasing ECF. On the other hand, $V_{OC}$ also increases with increasing UCW, which might be a bit surprising. This effect is a result of the increased distance between the closest points on two neighboring electrodes rather than the actual broadening of the doped regions. With applied voltage there will be a increased amount of lateral current in the highly doped regions near the rear-side. Longer distance between the electrodes will result in higher total recombination and a
higher voltage has to be applied to outweigh the photocurrent. Thus, an increased \( V_{\text{OC}} \). Further research remains to thoroughly understand this effect.

![Figure 5](image1.png)

**Figure 5:** The figure shows the open circuit voltage as a function of the width of the unit cell for different emitter coverage fractions.

Figure 6 shows the limiting efficiency of the interdigitated back-junction cell as a function of the ECF for different UCWs. The efficiency is increasing with increasing coverage fraction for all UCWs. A maximum efficiency of 21.7% is obtained with a unit cell width of 200-500 \( \mu \text{m} \). The increased efficiency with increasing coverage fraction results mainly because of the increased short-circuit current density.

![Figure 6](image2.png)

**Figure 6:** The figure shows the limiting cell efficiency as a function of the rear-side emitter coverage fraction for different unit cell widths. The remaining area of the back-side is assumed covered by BSF.

Our simulations indicate that a small unit cell is not always preferable. Figure 6 shows that the limiting efficiency decreases when decreasing the unit cell width to 100 \( \mu \text{m} \) at 80% emitter coverage fraction. This happens because the loss in open-circuit voltage in the smallest unit cell outweighs the moderate increase in short circuit current and fill-factor. The effect is only seen for high minority carrier lifetime materials. In materials with shorter diffusion lengths, we will benefit more from reducing the UCW because of the increased importance of collection efficiency and at the same the loss in \( V_{\text{OC}} \) is more moderate. \( V_{\text{OC}} \) drops from 0.670 V to 0.654 V when going from a 500 \( \mu \text{m} \) to a 100 \( \mu \text{m} \) wide unit cell with an effective lifetime of 2 ms. At the same time, \( V_{\text{OC}} \) only drops 3 mV (0.590 V to 0.587 V) at an effective MCL of 62 \( \mu \text{s} \). At small UCWs this reduction is large enough so that the limiting efficiency actually decreases slightly, despite of the increased short-circuit current.

5 CONCLUSIONS

These preliminary simulations have given us several interesting results regarding the design rules of our back-junction solar cell. We have found that the optimal cell thickness varies with effective minority carrier lifetime. This lifetime will in turn depend on the quality of the wafers and the passivation of both front and back-surfaces. The emitter should cover a large fraction of the back surface in order to maximize the collection of minority carriers and hence the short circuit current density. The open circuit voltage also increases with increasing emitter fraction. Our simulations show that 70-80% p+ emitter coverage fraction is preferable. The importance of increasing the coverage fraction increases with larger unit cells and shorter minority carrier diffusion lengths. For short effective lifetimes the unit cell width should be as small as possible to optimize the limiting cell efficiency. For high quality wafers with excellent surface passivation larger unit cells will give better results with our design because of an increased open circuit voltage.

6 REFERENCES


Paper II

LASER ABLATION OF PECVD OXIDE FOR STRUCTURING OF BACK-JUNCTION INTERDIGITATED SILICON SOLAR CELLS

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ABSTRACT: We use a Q-switched Nd:YVO₄ 532 nm laser with a spot size of 20 µm and ns pulses to locally ablate a layer of PECVD SiO₂ to structure the rear of an n-type back-contacted back-junction solar cell with interdigitated fingers. The PECVD SiO₂ works as a barrier against diffusion from the spray-on dopants used in the solar cell fabrication process, as well as a mask for an aqueous KOH solution that is used to remove the laser damage. The laser ablation energy threshold is found to be 0.1 mJ, which corresponds to a peak power density of 850 MW/cm². With optimized laser parameters, we show that an optimized KOH etch together with a set of laser parameters will increase the lifetime of the laser ablated areas to the same level as the untreated areas of the samples. It is also shown that there is a strong correlation between the surface roughness and the effective lifetime in the laser treated areas after the removal of the laser damage.

Keywords: Laser processing, silicon, high efficiency

1. INTRODUCTION

Back-contacted solar cells have served as an alternative to conventional solar cells for more than 30 years [1], and many different cells with quite different designs, such as metallization-wrap-through, emitter-wrap-through, and back-junction cells, have been developed [2]. Back-junction solar cells have both the emitter and the complete metallization on the rear side. This eliminates the shading present in conventional solar cells, and the sunward surface and back-side of the solar cell can therefore be independently optimized for optical and electrical performance, respectively. Moreover, having both contacts on the back side may significantly simplify module assembly, and allow for an increased packing density of the cells. This type of cell design requires high quality multicrystalline silicon wafers with long minority carrier lifetimes and excellent front surface passivation, since the carriers, created predominantly near the front surface, have to diffuse through the whole wafer. One of the main challenges in the processing of back-contacted interdigitated solar cells is the separation of the highly doped p- and n-fingers on the back of the cell. One possible way of doing this is to locally ablate a layer of diffusion barrier by laser. With lasers we have the possibility of a reproducible structuring of the cell at a speed very much compatible with in-line production. In these experiments we have investigated the effect on the minority carrier lifetime from ablation of a layer of PECVD SiO₂ with a Q-switched Nd:YVO₄ 532 nm laser with a spot size of 20 µm and a pulse width within the ns-range. Lasers with ns-pulses will induce damage in the underlying silicon due to thermal dissipation. This damage has to be removed in order to obtain high minority carrier lifetimes, which is a necessity in back-junction solar cells.

2. EXPERIMENTS

2.1 Preliminary investigations

Preliminary to the experiments the laser ablation energy threshold of the PECVD SiO₂ was investigated and found to be 0.1 mJ per pulse, corresponding to a peak power density of 850 MW/cm². The oxide was also found to work as a barrier against diffusion from the spray-on dopants used in our laboratory. The etch rate of the PECVD SiO₂ in KOH varied with concentration and temperature, but did not exceed 10 nm/min for any experiment. Thickness measurements were performed with a variable angle spectroscopic ellipsometer.

2.2 Preparation of samples

N-type (100) 1-3 Ωcm double side polished Cz wafers were used in all experiments. The native oxide was removed in a 1 min dip in 5% HF before a 500 nm thick layer of oxide was deposited on both sides of the wafer with a Plasmalab 133 Oxford PECVD system. The wafers were then cut into four equal samples. On each sample six 1 cm² squares of oxide were ablated together with some of the underlying silicon using different laser parameters. The laser spot to spot overlap was varied between 0 and 50%, while the diode current was tested at 27 A and 28 A, just above the ablation threshold. The pulse frequency was held fixed at 20 kHz. All samples were then etched in an aqueous KOH solution from 0 to 20 minutes with the PECVD oxide serving as an etching mask. KOH concentrations between 10% and 55%, by weight, were tested, and the temperature was altered between 68°C and 80°C. After KOH etching the PECVD oxide was stripped off in 5% HF, before the samples were thoroughly cleaned with Piranha (1:4 H₂O₂:H₂SO₄), HF, and HCl [3]. Finally, for passivation, a thin layer of PECVD a-Si:H was deposited on both sides of the samples before they were annealed in a belt furnace at 450°C for two minutes.

2.3 Measurements

A Semilab WC-2000 µW-PCD system was used to obtain a spatially resolved minority carrier lifetime map of the different samples. An example is shown in Fig. 1. The surface roughness of the laser ablated areas was measured, before and after KOH etching, by an Alpha-step profilometer by scanning orthogonally to the laser scanning direction. A typical scan is shown in Fig. 2.
Figure 1: A spatially resolved minority carrier lifetime ratio map from a Semilab WC-2000 µW-PCD system. Four different samples, cut from the same wafer, with different KOH etching times are shown, put back together. The etching time is increasing from the upper left sample, to the lower left, and further from the upper right to the lower right sample. The lifetime is compared to the average lifetime of the untreated areas.

3. RESULTS

3.1 Minority carrier lifetime

The minority carrier lifetime is very poor just after laser ablation for all the different laser parameters. The laser, with pulses in the ns-range, will induce thermal damage in the underlying silicon lattice. The reach of the thermal dissipation can be approximated by

\[ I_T = 2\sqrt{Dt} \]

where \( I_T \) is the thermal penetration depth, \( D \) is the silicon thermal diffusion coefficient, and \( t \) is the pulse length.

Figure 2: A profilometer scan of a KOH etched laser ablated area together with some of the untreated area. The scanning is done orthogonally to the laser scanning direction.

For a 75 ns laser pulse this implies that at least 5 µm of silicon should be etched away before the thermal damage is removed. The time it takes to remove the damage is therefore dependent of the KOH etching rate, which is a function of both concentration and temperature [5]. In Fig. 3 the minority carrier lifetime ratio is plotted as a function of a etching time for a 10% (wt) concentration, 68°C KOH etch for different sets of laser parameters. The lifetime increases with the etching time from the start, as expected, but then starts to decrease again after a certain time. This is a consequence of the low KOH concentration and the laser parameters used for this data set. This problem can be omitted by using a higher concentration KOH solution and an appropriate set of laser parameters. This will be shown in the next section.

3.2 Surface Roughness

The etched laser pits should have as smooth a surface as possible to reduce the surface area and the number of recombinative surface states and, thus, the surface recombination. We calculate the surface roughness, \( R_s \), as the arithmetic average of absolute amplitudes. \( R_s \) is very sensitive to the laser parameters and the KOH etching time, temperature, and concentration. KOH concentrations lower than 20% will result in hillock formation [6], and, thus, an increase in the surface roughness with time. In Fig. 4 we can see how the surface roughness first decreases with time, but then starts to increase again for a low concentration KOH etch and non-optimized laser parameters. The correlation between the surface roughness and the minority carrier lifetime is clearly seen from the figure.

Figure 3: Lifetime ratio vs etching time for a 10% (wt) concentration, 68°C KOH solution. The spot to spot overlap was varied between 0 and 50%, while the diode current and pulse frequency were 28 A and 20 kHz, respectively. The lifetime is compared to the average lifetime of the untreated areas.
Figure 4: Minority carrier lifetime ratio and surface roughness as function of etching time for a 10% (wt) concentration, 68°C, KOH solution. The spot to spot laser overlap was 20%, while the diode current and the pulse frequency were 28 A and 20 kHz, respectively. The benefit of applying appropriate laser parameters and a high concentration KOH etch can be seen in Fig. 5. The minority carrier lifetime increases rather quickly with time and it does not have a significant drop after maximum lifetime is reached. Neither does the surface roughness increase with time. Our best resulting surface roughness for the different KOH concentrations and temperatures is comparable to that of Noor et al [7], who studied KOH etching on smooth surfaces.

Figure 5: Minority carrier lifetime ratio and surface roughness as function of etching time for a 55% (wt) concentration, 80°C, KOH solution. The spot to spot laser overlap was 50%, while the diode current and the pulse frequency were 27 A and 20 kHz, respectively. The lifetime is compared to the average lifetime of the untreated areas.

4. CONCLUSION

The PECVD SiO\(_2\) layer serves as both a diffusion barrier and as an etching mask against an aqueous KOH solution. With the use of an appropriate set of laser parameters and a high concentration KOH etch the damage induced in the silicon due to thermal dissipation from laser during the SiO\(_2\) ablation, can be completely removed. The surface will be more or less smooth, and the minority carrier lifetime in the laser ablated areas will be in the same range as in the untreated areas.

REFERENCES

Paper III

K. Mangersnes, S. E. Foss, and A. Thøgersen. "Damage free laser ablation of SiO₂ for local contact opening on silicon solar cells using an a-Si:H buffer layer."

Damage free laser ablation of SiO$_2$ for local contact opening on silicon solar cells using an a-Si:H buffer layer

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We have used a Q-switched Nd:YVO$_4$, diode pumped 532 nm laser with nanosecond pulses, and a spot diameter of 40 mm to ablate a layer of plasma enhanced chemical vapor deposited (PECVD) SiO$_2$ on n-type Cz silicon, with the aim of making local contact openings on back-junction silicon solar cells. Laser pulses within the ns range are usually believed to be incompatible with processing of high efficiency solar cells because such long pulses induce too much damage into the underlying silicon lattice. This is due to thermal dissipation. In this work, a PECVD layer of a-Si:H between the n-type silicon and the dielectric layer is shown to absorb much of the laser radiation and allows for ablation at laser fluxes lower than the ablation threshold of crystalline silicon. In addition, the a-Si:H layer serves as an excellent surface passivation layer for the silicon substrate. We show that it is possible to ablate PECVD SiO$_2$ in a damage free way with fluxes five times lower than those needed to ablate crystalline Si. Our results are verified experimentally with high resolution transmission electron microscopy of the crystal structure in the laser irradiated areas, and quasi-steady-state photoconductance measurements of emitter saturation currents. In addition, we have simulated the energy transfer from a ns 532 nm Gaussian shaped laser beam to a SiO$_2$ covered Si lattice with and without the a-Si:H buffer layer. A model that coincides very well with the experiments is found. © 2010 American Institute of Physics. [doi:10.1063/1.3309382]

I. INTRODUCTION

Back-contacted silicon solar cells have served as an alternative to conventional solar cells for more than 30 years. Several different solar cell designs, such as metallization-wrap-through, emitter-wrap-through, and back-junction cells, have been developed. Back-junction solar cells have both the emitter and the complete metallization on the rear side. This eliminates the contact shading present in conventional solar cells. The sunward surface and back-side of the solar cell can therefore be independently optimized for optical and electrical performance, respectively. Moreover, having both contacts on the back side may significantly simplify module assembly, and allow for an increased packing density of the cells. Back-junction solar cells require high quality monocrystalline silicon wafers with long minority carrier lifetimes and excellent surface passivation. Low surface recombination losses in silicon solar cells using an a-Si:H buffer layer have been developed. a-Si:H layer serves as an excellent surface passivation layer for the silicon substrate. We show that the thermal diffusion can be approximated by

\[ I_d = 2\sqrt{\alpha t}, \]

where $I_d$ is the thermal penetration depth, $\alpha$ is the thermal diffusivity, and $t$ is the pulse length of the laser. In silicon, the thermal penetration depth of a 100 ns long laser pulse will be around 3 $\mu$m at 1000 K. In this paper, we show that local contact can be made with a ns laser without damaging the crystalline silicon, and thus maintain a high minority carrier lifetime, which is a necessity in high efficiency silicon solar cells. Our solution is to add a buffer layer of plasma enhanced chemical vapor deposited (PECVD) a-Si:H (a-Si for simplicity) between the oxide and the silicon, and utilize the thermal diffusivity of a-Si, which is two orders of magnitude lower than that of silicon, to isolate the laser damage. The higher optical absorption coefficient of a-Si compared to crystalline silicon also contributes to the concentration of laser energy within the a-Si buffer layer.

II. EXPERIMENTAL

n-type Cz 1–3 $\Omega$ cm Si wafers were used in all the experiments. Before processing, the wafers were cleaned in piranha ($H_2SO_4:H_2O_2$, 4:1) and HF (5%).

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A. Laser ablation energy threshold

We prepared one set of wafers for laser fluence ablation threshold experiments. Laser fluence is the pulse energy density, and it is given in units of J/cm². We refer in the following to the laser fluence at the peak of the Gaussian beam profile. The laser we used was a Q-switched Nd:YVO₄, diode pumped 532 nm laser with nanosecond pulses, and a 1/e² spot diameter of 40 μm. For these experiments, the wafers were cut into four parts. Thereafter, a PECVD layer of a-Si of varying thickness (0–1000 nm) was deposited on the different samples. The samples were then covered with 500 nm of PECVD SiO₂. The thickness of the a-Si layer was measured by a variable angle spectroscopic ellipsometer. Each sample was exposed to the same matrix of single laser pulses, with varying fluence in the range 0.1–2 J/cm². In this energy range, the pulse duration decreases close to linearly with increasing fluence, from 150 ns at 0.1 J/cm² to 80 ns at 2 J/cm². We then determined the ablation threshold of the oxide by visual inspection in an optical microscope.

B. Local emitter saturation current

A set of four wafers was prepared for phosphorus doping. Each wafer was doped on both sides by spraying on a diluted phosphorus containing dopant source with a custom-built spray-on system and subsequent drying and indiffusion in a belt furnace. The phosphorus source was P509 from Filmtronic. Similarly, four wafers were boron-doped with boron-A from Filmtronic as a dopant source. A third set of four wafers was left undoped. The diffused phosphorus layers had a sheet resistivity of 50±5 Ω/□ (measured on a p-type wafer added for monitoring the sheet resistance), while the more shallow boron emitters had sheet resistivities of 150±20 Ω/□. After the diffusion, the phosphorus and boron containing glass layers were removed in HF (5%) and in a belt furnace. The required back side surface passivation for lifetime measurements was obtained by depositing a thin layer of a-Si on the back side of the samples by parallel plate PECVD in an Oxford Plasmalab 133 system. The amorphous layer was deposited at 230 °C with a flow of 25 SCCM (SCCM denotes standard cubic centimeter per minute at STP) of undiluted silane (SiH₄). On two wafers of each set, nondiffused, phosphorus diffused, and boron-diffused, a thicker layer of a-Si was deposited on the front side. All the wafers were then covered with 500 nm of PECVD SiO₂ and cut into four samples of equal size. The different samples were exposed to a matrix of single laser pulses, covering approximately 20% of the sample area. The laser fluence was varied from 0.2 to 1.4 J/cm². After the laser ablation, the SiO₂ was stripped off in HF (5%). Thereafter, a new thin layer of a-Si was deposited on the front of all samples, using the same method as before. Finally, all samples were annealed for 2 min at 450 °C in a belt furnace.

We used quasi-steady-state photoconductance technique measurements under high injection conditions to extract the local emitter saturation current. The effective saturation current before and after laser ablation was used together with the fraction, a, of the laser ablated area to the total sample area to calculate the emitter saturation current in the laser ablated spots. $J_{oe, eff} = a \times J_{oe} + (1-a) \times J_{oe, pass}$, where $J_{oe, eff}$ is the effective emitter saturation current, $J_{oe}$ is the local emitter saturation current in the laser ablated spots, and $J_{oe, pass}$ is the emitter saturation current of the passivated sample before laser ablation. This gives $J_{oe} = a^{-1} [J_{oe, eff} - (1-a) \times J_{oe, pass}]$ for the local emitter saturation current in the laser radiated spots. A similar analysis was done by Engelhart et al. when comparing contact openings made by ns and picosecond lasers.

C. Transmission electron microscopy

High resolution transmission electron microscopy (HR-TEM) was used to analyze the defects induced in the crystal lattice from the laser ablation at different laser energies with and without an a-Si buffer layer of 500 nm. We used a 200 keV JEOL 2100F microscope equipped with a Gatan imaging filter and detector. Cross sectional samples were prepared by gluing two sample substrates together, front to front, with epoxy glue. The cross sectional samples were thinned using ion-milling with a Gatan precision ion-polishing system operated at 5 kV gun voltage. Figure 1 shows the laser ablation process sequence, and the cross section area used for TEM analysis.

III. SIMULATIONS

We used the ATLAS device simulator, within the Silvaco simulation framework, to simulate the energy transfer from a laser beam to a SiO₂ covered silicon substrate with and without a 500 nm thick a-Si buffer layer between the oxide and the silicon. The simulations were done to illustrate the general idea presented herein, of using an optically and thermally absorbing buffer layer for avoiding defect formation in the silicon substrates. They do not, however, give an exact measure of any physical quantity. In the simulations, a 532 nm Gaussian shaped laser beam, with a pulse duration of 90 ns was directed normally toward the surface of the two different samples. The ramping time of the beam was set to 100 ps. A low beam power was used to avoid temperatures near the melting point of a-Si and Si, as the device simulator is not meant to handle phase transitions. The heat capacity of the a-Si layer was modeled according to Endo et al. as

$$C_p = 13.1 + 2.25 \times 10^{-2} T - 8.3 \times 10^{-6} T^2 \text{ (Jmol}^{-1} \text{K}^{-1}),$$

where $C_p$ is the molar heat capacity and $T$ is the temperature ranging from 300 to 1000 K. The heat conductivity of a-Si...
was assumed to be a constant value of 0.010 W/cm·K.\textsuperscript{14,23} The simulated heat distribution after one single laser pulse for the two samples without and with the a-Si layer is shown in Figs. 2 and 3, respectively. These figures clearly show that the a-Si layer concentrates the laser energy both optically and thermally, thereby confining the damage to a smaller volume.

IV. RESULTS AND DISCUSSION

A. Laser ablation energy threshold

The fluence needed to ablate the PECVD SiO\textsubscript{2} on bare silicon was determined experimentally to be 1.5 J/cm\textsuperscript{2}, but with a 500 nm thick layer of a-Si between the silicon and the oxide the ablation fluence threshold was reduced to a much lower value of 0.3 J/cm\textsuperscript{2}. We found that the ablation threshold decreases with an increasing thickness of the a-Si layer, as shown in Fig. 4. Already at an a-Si thickness of 25 nm, the fluence ablation threshold is reduced to less than one third of the fluence needed to ablate SiO\textsubscript{2} on crystalline silicon. For a-Si films thinner than approximately 40 nm, there is an exponential decay in the laser fluence ablation threshold. We believe that the exponentially decreasing ablation threshold with increasing film thickness is a result of more energy being optically absorbed in the a-Si film rather than being transmitted through to the silicon substrate where the energy will be less spatially confined. For thicker films, where the optical absorption is no longer a limiting factor, we find a linear decrease in ablation threshold, before the laser fluence ablation threshold reaches a constant value for films thicker than approximately 400 nm. The onset of the constant fluence ablation threshold is equal to our estimated thermal diffusion length of a-Si. A thermal diffusion length of 400 nm corresponds to a thermal diffusivity of 3.5 × 10\textsuperscript{-3} cm\textsuperscript{2}/s, two orders of magnitude less than that of crystalline silicon.\textsuperscript{14} The laser fluence ablation threshold for thin metal films on fused silica was studied by Matthias et al.,\textsuperscript{24} who also developed a thermodynamic model for the fluence ablation threshold of the thin film. They found a linear and a constant region, both experimentally and theoretically, but in their case with a linear decrease in the threshold fluence with increasing film thickness. The energy density in the metal film decreases with increasing film thickness since the substrate is a poor thermal conductor compared to the metal film. The fluence ablation threshold was found to be constant for films equal to or thicker than the thermal diffusion length of the metal film. In our case, the substrate is a better thermal conductor than the film, and more heat will dissipate into the substrate as the film gets thinner. This is the reason why we have opposite linear dependence in the two cases. The exponentially decreasing region was not observed by Matthias et al.\textsuperscript{24} due to the very short optical penetration depth of the metals studied, and the short wavelength of the laser, but it was discussed that this region should exist only if the model would be valid. Figure 5 shows an optical microscope image of contact openings locally ablated with the ns laser, at fluences just above the ablation threshold of the a-Si layer. It should also be noted that we have tested the variation in the laser fluence ablation threshold energy with varying annealing times and temperatures of the a-Si layer. For temperatures up to 500 °C only very small changes in the ablation threshold energy was found. We, therefore, believe...
that the amount of hydrogen in the a-Si layer is not important for the laser fluence ablation threshold energy.

B. Local emitter saturation current

The emitter saturation current gives an indication of the electronic quality of the emitter. It is a measure of the total recombination taking place in the emitter, also including surface recombination, recombination at laser induced defects, and band gap narrowing.\(^{23-25}\) For a diffused emitter the saturation current can be given in terms of the effective surface recombination velocity through

\[
S_{\text{eff}} = \frac{J_{\text{oc,eff}} N_A}{n_i^2},
\]

where \(S_{\text{eff}}\) is the effective surface recombination velocity, \(J_{\text{oc,eff}}\) the emitter saturation current, \(N_A\) is the doping level of the diffused emitter, and \(n_i\) the intrinsic carrier density.\(^{7}\) In the absence of a diffused emitter, as for the undoped samples, \(J_{\text{oc,eff}}\) will give an indication of the surface passivation quality. Figure 6 shows the local emitter saturation current and the local saturation current for the doped and undoped samples, respectively, as a function of the laser fluence for the samples prepared without the extra a-Si layer. For the undiffused samples \(J_o\) increases with increasing laser fluence, indicating that the laser damage reduces the minority carrier lifetime. However, for small laser fluences \(J_o\) seems more or less unaffected by the laser pulse. The same trend is also seen for the shallow (150 Ω\(\cdot\)cm) boron emitter. For the deeper phosphorus emitter (50 Ω\(\cdot\)cm), no effect of the laser pulse can be seen from the figure. The fluctuation in \(J_o\) observed in Fig. 6 is most likely to result from a small spread in sheet resistivity over the wafer. The effect of increasing emitter damage is less pronounced for the samples prepared with a buffer layer of a-Si. This is shown in Fig. 7. In the low laser fluence region, near the a-Si ablation threshold, \(J_o\) of the undiffused samples seems to be more or less unaffected by the laser ablation. It is also worth noting that the reference value of \(J_o\) at zero fluence for the undiffused samples is lower than the reference value of the corresponding samples without the buffer layer. We believe that this is because the relatively thick buffer layer of a-Si passivates the surface even better than the thin layer used for passivation. The threshold laser fluence where laser damage occurs is shifted to higher fluences compared to the experiments without an a-Si layer. The same can be seen for the shallow boron emitters. As for the phosphorus doped samples, there seems to be little influence from the laser irradiation on the local emitter saturation current.

C. Damage to the crystal structure

Figure 8 shows a HRTEM picture of a sample that was prepared with 500 nm a-Si under the oxide layer. The sample was exposed to single laser pulses with fluences just above the minimum ablation threshold of SiO\(_2\) on a-Si. The SiO\(_2\) is stripped off, as before, but it can also be seen that the a-Si layer has been locally removed. Still, there is no visible damage to the crystalline silicon. Several HRTEM pictures were taken, and no visible damage could be found on this sample. This was also indicated by the emitter saturation current measurements, showing no increase with low laser fluences compared to the nonlaser processed reference. We found that a laser fluence of 0.3 J/cm\(^2\), which corresponds to the minimum ablation threshold fluence of SiO\(_2\) on a-Si, is not enough to cause visible damage to the crystal lattice even without the a-Si buffer layer. This implies that we are able to make local contact through an a-Si–SiO\(_2\) stack without damaging the underlying crystalline silicon. A more thorough analysis of the HRTEM induced defects will be presented.

![Image](jap.aip.org/jap/fig/5.png)

FIG. 5. (Color) An optical microscope picture of local contact openings of SiO\(_2\) ablated together with a buffer layer of amorphous silicon on a crystalline silicon sample. The laser fluence used to make the openings was about 0.3 J/cm\(^2\); five times lower then the laser fluence needed to ablate crystalline silicon. The diameter of the ablated spots is 20 μm.

![Image](jap.aip.org/jap/fig/6.png)

FIG. 6. (Color online) Local emitter saturation current density and local saturation current density for the doped and undoped samples, respectively, as a function of laser fluence for samples without an a-Si buffer layer, shown for undiffused samples (circles), P-diffused samples (triangles), and B-diffused samples (squares). The reference values for the nonablated samples are shown at zero fluence.

![Image](jap.aip.org/jap/fig/7.png)

FIG. 7. (Color online) Local emitter saturation current density and local saturation current density for the doped and undoped samples, respectively, as a function of laser fluence for samples with a 500 nm thick a-Si buffer layer, shown for undiffused samples (circles), P-diffused samples (triangles), and B-diffused samples (squares). The reference values for the nonablated samples are shown at zero fluence.
a-Si have been ablated with a single laser pulse. The fluence used was 0.3 J/cm². a-Si have been ablated with a single laser pulse without any damage to the silicon crystal structure. The fluence used was 0.8 J/cm². The sample was exposed to single laser pulses. Each with a fluence of 0.8 J/cm², and the damage to the crystal structure is easily seen from the picture. The damage origins from a specific depth of 26 nm under the silicon surface. Several HRTEM pictures were taken on this sample, and the same defects at the same specific depth were found on every picture. The details of the HRTEM experiments will be presented in an upcoming article.¹²

V. CONCLUSION

We have shown that it is possible to locally ablate a dielectric layer on silicon in a damage free way with the use of a long pulsed laser by adding a buffer layer of a-Si between the dielectric and the silicon. The a-Si has a larger optical absorption coefficient, and the thermal diffusivity is more than two orders of magnitude lower than that of crystalline silicon. Therefore, the laser energy will be confined to a much smaller volume, and the ablation takes place at significantly lower fluences. The laser fluence needed to ablate a-Si film decreases with increasing film thickness up to a thickness of 400 nm, corresponding to our estimated thermal penetration depth of a-Si, where the threshold energy reaches its minimum value of 0.3 J/cm². a-Si is also known to be an excellent surface passivation layer for both p- and n-type silicon solar cells. We believe that this general idea, of introducing an absorbing and insulating a-Si buffer layer, will be even more relevant for lasers with shorter pulse durations, as the thickness of the a-Si layer and thereby the deposition time can be significantly reduced.

ACKNOWLEDGMENTS

The authors would like to thank Gunnar Rustad, Øystein Farsund, and Magnus W. Haakested at the Norwegian Defense Research Establishment for valuable help and advice regarding measurements of the laser characteristics.

²⁶ P. Engel hard, R. Grischke, S. Eidelloth, R. Meyer, A. Schoonder beek, U.
Paper IV

A thermodynamic model for the laser fluence ablation threshold of PECVD SiO$_2$ on thin a-Si:H films deposited on crystalline silicon

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ABSTRACT

We have developed a thermodynamic model that predicts the heat distribution in a stack of PECVD SiO$_2$ and a-Si:H on crystalline Si after laser irradiation. The model is based on solving the total enthalpy heat equation with a finite difference scheme. The laser used in the model is a frequency doubled Nd:YVO$_4$ green laser with pulse duration in the nanosecond range. The modeling was done with the aim of getting a better understanding of our newly developed laser ablation process for making local contacts on back-junction silicon solar cells. Lasers with pulse duration within the nanosecond range are usually believed to induce too much thermal damage into the underlying silicon to make them suitable for high efficiency solar cells. In our case, insertion of a thin layer of a-Si:H between the SiO$_2$ and the Si absorbs much of the laser irradiation both optically and thermally. This makes it possible to form local contacts to Si in a damage-free way. In addition, the residual a-Si:H serves as an excellent surface passivation layer for the Si substrate. We have also developed a simple static model to determine the onset of SiO$_2$ ablation on a-Si:H layers of varying thickness. The models, both the static and the dynamic, are in good agreement with experimental data.

INTRODUCTION

Lasers have in the recent years been shown to be a very promising tool for making local contacts on silicon solar cells through a dielectric layer[1-7]. The local ablation is done to reduce the total metal semiconductor area of the solar cells, and thus improve the efficiency of the device. Lasers in the nanosecond (ns) range are usually believed to induce too much thermal damage into the underlying silicon lattice to make them suitable for high efficiency solar cell concepts[4, 5, 8]. In the case of a frequency doubled Nd:YVO$_4$ green laser, SiO$_2$ is transparent to the laser light. Ablation of SiO$_2$ on Si takes place through an indirect process where the oxide is lifted off from expansion of molten or vaporized silicon. We have recently shown that insertion of a buffer layer of amorphous silicon, a-Si:H (a-Si for simplicity), between the SiO$_2$ and the Si will absorb much of the laser irradiation[7]. The high optical absorption and low thermal conductivity of a-Si compared to Si confines the laser energy to a much smaller volume. This makes it possible to ablate SiO$_2$ in a damage free way with a laser fluence five times lower than that needed to ablate SiO$_2$ on crystalline silicon. In addition, the residual a-Si serves as an excellent surface passivation layer for the Si substrate[9-11]. a-Si is not a very well defined material as most of the material parameters are very sensitive to the deposition technique and parameters. Especially the absorption coefficient and the thermal conductivity vary over a broad range according to different references[12-19]. Previous work on laser interaction with a-Si has mostly focused on melting and recrystallization of a-Si, both experimentally[15, 20-22] and numerically[21, 23-25]. The goal of this paper is to develop a thermodynamic model that predicts the laser fluence ablation threshold of SiO$_2$ on a-Si layers of varying thickness, and that also describes the heat distribution in the SiO$_2$ - a-Si - Si stack after laser irradiation.
THEORY

To find an expression for the laser fluence ablation threshold we make some simple assumptions. We assume that the onset of melting or vaporization is given by a critical energy density \( E_T = F_T / L_{th, eff} \), where \( F_T \) is the threshold fluence, \( L_{th, eff} \) is the effective thermal diffusion length defined by \( L_{th} = \sqrt{2 D_{eff} \tau_p} \), where \( D_{eff} \) is the effective thermal diffusivity and \( \tau_p \) is the pulse duration of the laser at full width half maximum. \( D_{eff} \) is the ratio between the effective thermal conductivity, \( k_{eff} \), and the volumetric heat capacity. \( k_{eff} \) is weighted between the thermal conductivity of a-Si and Si, also including a thermal resistance at the interface between the two. A version of this model, not including interface resistance, was proposed by Matthias et al., when investigating laser ablation of metal films on quartz[26]. Herein we also need to account for an optically absorbing substrate under the thin film. If we assume a uniform temperature rise throughout the volume defined by \( L_{th, eff} \), it can be shown that the threshold fluence for melting is given by

\[
F_{TM} = \frac{\Delta T_M}{(1-R)(1-e^{-\alpha d})+(1-e^{-\beta d})e^{-\alpha d}} C_{p, eff} L_{th, eff}
\]

\( \Delta T_M \) is the needed temperature increment for melting, \( d \) and \( d_s \) are the thicknesses of the a-Si layer and the Si substrate, respectively. We assume that the transparent SiO\(_2\) constitutes a thermal barrier and only influences the reflection, \( R \), at the surface. \( \alpha \) and \( \beta \) are the optical absorption coefficients of a-Si and Si, respectively. \( C_{p, eff} \) is the effective volumetric heat capacity; linearly weighted between a-Si and Si. The corresponding model for the onset of surface vaporization is the same as for melting but with \( \Delta T_M \) replaced by \( \Delta T_V + \Delta H_m / C_{p, a-Si} \), where \( \Delta T_V \) is the needed temperature rise for vaporization, \( \Delta H_m \) is the latent heat of melting, and \( C_{p, a-Si} \) is the volumetric heat capacity of a-Si. The parameters used in the model are taken from Table I. We used the respective values at 1000 K for the temperature dependent parameters, and a pulse duration of 142 ns. Our experimental data from reference 7 is plotted together with the static thermodynamic models for surface melting and vaporization in Figure 1. There is a good fit between the experimental data and the static model for surface evaporation even without temperature dependent parameters, and with the assumption of a uniform temperature rise. Still, this model only gives an indication of the onset of evaporation and does not include any information about the actual temperature distribution within the film and the substrate. Neither have we taken into account that the pulse length of our laser varies with fluence as given in Table I. A more dynamic description that also includes phase transitions will require a solution of the total enthalpy version of the heat equation. The diameter of our laser spot is much wider than the thermal diffusion length of a-Si, and it is a good approximation to solve the one-dimensional version of the equation[27].

\[
\frac{\partial \Delta H(x,t)}{\partial t} = \frac{\partial}{\partial x} \kappa(T) \left( \frac{\partial}{\partial x} T(x,t) \right) + S(x,t)
\]
\( \kappa(T) \) is the temperature dependent thermal conductivity, \( x \) is the depth from the top of the a-Si layer, and \( S(x,t) \) is the laser source given by

\[
S(x,t) = \frac{4 \ln 2}{\pi} \frac{(1 - R)F \cdot \alpha}{\tau_p} \exp \left( -\alpha x - 4 \ln 2 - \frac{(t - t_{\text{peak}})^2}{\tau_p^2} \right)
\]  

(3)

\( F \) is the fluence at the peak of the Gaussian pulse, and \( t_{\text{peak}} \) is the time for the peak fluence. The total enthalpy, \( \Delta H \), is given by[27]

\[
\Delta H(T) = \int_{T_c}^{T} \rho(T') c_p(T')dT' + \eta(T - T_M)\Delta H_M + \eta(T - T_V)\Delta H_V
\]

(4)

\( T_0 \) is the ambient temperature, and \( T_M \) and \( T_V \) are the melting and vaporization temperatures, respectively. \( \eta \) is the Heaviside function (1 or 0 if the argument is positive or negative, respectively) and \( \Delta H_m \) and \( \Delta H_v \) are the latent heat of melting and vaporization, respectively.

**Figure 1**: The fluence threshold for laser ablation as a function of the thickness of the a-Si film. Our experimental data from reference 7 (solid line with circles) is shown together with the modeled data for surface melting and vaporization.

To omit numerical instabilities we have assumed that the phase transitions occur over a temperature interval of \( \Delta T = 5 \) K. The SiO\(_2\) is assumed to be optically transparent and thermally insulating, and is only influencing the surface reflection. The structure is at ambient temperature before the laser heating starts. We have used the following initial and boundary conditions:

\[
T(x,t) = T_0 \bigg|_{x=0} \quad T(x,t) = T_0 \bigg|_{x=\infty} \quad \frac{\partial T(x,t)}{\partial x} = 0 \bigg|_{x=0}
\]

(5)

A finite difference scheme was used to solve the total enthalpy heat equation. The simulation parameters are listed in Table I. a, c, and l in the table refer to amorphous Si, crystalline Si, and liquid Si, respectively.
DISCUSSION

In Figure 2 we show the temperature distribution in a SiO$_2$ covered silicon wafer at different depths after laser irradiation with a single laser pulse with a fluence of 0.32 J/cm$^2$. In Figure 3 we show the temperature distribution in a similar sample, but now with a 300 nm buffer layer of a-Si. We observe that the a-Si obtains a much higher surface temperature, and that 0.32 J/cm$^2$ is enough to reach the onset of surface vaporization. This result, and also results for a-Si layers of different thicknesses (not shown) are in good agreement with our experimental data.

<table>
<thead>
<tr>
<th>TABLE I</th>
<th>Value/Expression</th>
<th>Reference</th>
</tr>
</thead>
<tbody>
<tr>
<td>Absorption coefficient (cm$^{-1}$)</td>
<td>a: $1.9 \times 10^5$, c: $5.02 \times 10^3 e^{1.43T}$, l: $1 \times 10^6$</td>
<td>28, 20, 23</td>
</tr>
<tr>
<td>Density (g/cm$^3$)</td>
<td>a: 2.2, c: 2.32, l: 2.52</td>
<td>23, 27, 27</td>
</tr>
<tr>
<td>Thermal Conductivity (W/cm K)</td>
<td>a: $1.3 \times 10^{-1} \times (T-900)^4 + 1.3 \times 10^{-6} \times (T-900)^2 + 1 \times 10^{-8} \times (T-900) + 1 \times 10^{-2}$, c: $1521/T^{1.226}$ ($T&lt;1200$ K), $8.96/T^{0.502}$ ($1200$ K $&lt;=$ $T &lt; 1690$ K), l: 0.62</td>
<td>20, 20, 29</td>
</tr>
<tr>
<td>Heat capacity (J/cm$^3$K)</td>
<td>a: $2.2 \times (0.952 + 0.171 \times T/685)$, c: $2.32 \times (0.711 + 0.255 \times (T^{1.85-1})(T^{1.85-1} + 0.255/0.711))$, l: 1.0</td>
<td>20, 30, 29</td>
</tr>
<tr>
<td>Melting temp (K)</td>
<td>a: 1420, c: 1690</td>
<td>27, 27</td>
</tr>
<tr>
<td>Vaporization temp. (K)</td>
<td>l: 2680</td>
<td>27</td>
</tr>
<tr>
<td>Interface thermal resistance (cm$^2$ K/W)</td>
<td>0.0054</td>
<td>16</td>
</tr>
<tr>
<td>Latent heat of melting (J/K)</td>
<td>a: 1250, c: 1780</td>
<td>27, 27</td>
</tr>
<tr>
<td>Latent heat of Vaporization (J/K)</td>
<td>l: 15000</td>
<td>27</td>
</tr>
<tr>
<td>Reflection</td>
<td>a: 0.18, c: 0.18, l: 0.07</td>
<td>Measured, Measured, Calculated with data from ref 31</td>
</tr>
<tr>
<td>Pulse duration, full width at half maximum (ns)</td>
<td>-37F + 154</td>
<td>Measured, F is the peak fluence in the Gaussian beam profile (J/cm$^2$)</td>
</tr>
<tr>
<td>Pulse peak (ns)</td>
<td>200</td>
<td>Set</td>
</tr>
</tbody>
</table>

**Interface thermal resistance**

Kuo et al. measured the thermal resistance of the interface between a-Si films and crystalline Si[16]. The value we have used in our simulations, 0.54 mm$^2$ K/ W, is the resistance
they measured for a structure with a native oxide between the a-Si and the Si. This value was found to give the best fit between the modeling and our experiments. A clean surface, prepared by pre-sputtering in a vacuum chamber, showed an interface resistance of 0.15 mm² K/ W[16]. A lower interface resistance increases the laser fluence ablation threshold in our model, as less heat is contained in the film. The increased thermal resistance at the interface between a-Si and the Si (presumably provided by an oxide) is therefore a necessary element of our damage-free laser-ablation process.

CONCLUSIONS

We have developed a static thermodynamic model that describes the fluence threshold for pulsed laser ablation of SiO₂ on Si with a buffer layer of a-Si of varying thickness. We have also solved the heat equation for the SiO₂ – a-Si – Si stack to predict the temperature distribution within the a-Si film and the Si substrate. We found that an interface thermal resistance between a-Si and Si is important for lowering the laser fluence ablation threshold and for protecting the underlying silicon lattice. The results are in good agreement with our experimental data, when we include in the model the expected contribution from the native oxide layer to the thermal resistance of the interface.

Figure 2: Heat distribution in Si as a function of time at different depths after single pulse laser irradiation with a fluence of 0.32 J/cm² and a pulse duration of 142 ns. x = 0 (surface) corresponds to the top of the Si substrate.

Figure 3: Heat distribution as a function of time at different depths in a stack of 300 nm a-Si on Si after single pulse laser irradiation with a fluence of 0.32 J/cm² and a pulse duration of 142 ns. x = 0 (surface) corresponds to the top of the a-Si film.
REFERENCES

Paper V

Investigating the specific contact resistance on boron emitters in laser ablated holes in a passivating double-layer of a-Si:H and a-SiO$_x$:H

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*Institute for Energy Technology, Department of Solar Energy

(Dated: Sept 2010)

Abstract

We have investigated the specific contact resistance of sputtered aluminum contacts on boron emitters on n-type CZ Si wafers. The contacting was done through laser ablated holes in a stack of PECVD a-Si:H and a-SiO$_x$:H. The laser is a Nd:YVO$_4$, diode pumped, frequency doubled, green laser with pulse duration in the nanosecond range. Nanosecond lasers are usually believed to induce too much thermal damage into the underlying silicon lattice to be suitable for production of high efficiency silicon solar cells. Herein, the a-Si:H layer absorbs much of the laser irradiation both optically and thermally and makes it possible to make low resistance contacts even with a nanosecond laser. Specific contact resistance as low as 0.8 mΩ·cm$^2$ was measured for sputtered Aluminum. Finally, the impact of the contact resistance on the performance of a back-contact back-junction silicon solar cell was investigated through two-dimensional device simulations based on the finite element method.

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1. INTRODUCTION

Back-contacted silicon solar cells comprise solar cells where electrical contacts of both polarities are located on the backside. Several different back-contacted solar cell designs, such as emitter-wrap-through, metallization-wrap-through, and back-junction silicon solar cells have been developed during the last 35 years [1–3]. Back-contacted back-junction (BC-BJ) solar cells have both the emitter and the complete metallization on the rear side. This eliminates the contact shading present in conventional solar cells. The sunward and back surface can therefore be independently optimized for optical and electrical performance, respectively. Moreover, having contacts of both polarities on the same side may simplify module assembly and allows for an increased packing density of the cells[4–6]. Since the collecting emitter is located on the back of the cells, the BC-BJ cell design requires high-quality monocrystalline wafers with long minority carrier lifetimes and excellent surface passivation. Low surface recombination losses in silicon solar cells can be obtained by passivating the surface by SiO₂, SiNₓ, a-Si:H, Al₂O₃, and different stacks thereof[7–15], and by limiting the metal to semiconductor contact area. The metal-semiconductor contact surface is an area where the recombination of charge carriers is very high[16], and thus have a negative influence on the performance of a solar cell. This is mainly because of an increased saturation current and thereby a reduced open circuit voltage. By producing good quality, low resistive contacts, this area can be reduced without reaching critical resistance values, and thus increase the overall efficiency of the solar cell.

Lasers have in the recent years been shown to be a very promising tool for making local contact openings to silicon through a dielectric layer[17–20]. Solar cell concepts that earlier only have been possible to realize in low-throughput, high-cost lab facilities are expected to enter the commercial market in the coming years. In a recent paper, we showed how it is possible to locally ablate a double-layer film of plasma-enhanced chemical vapor deposited (PECVD) a-Si:H and a-SiO₂:H (a-Si and SiO₂, respectively, for simplicity) in a damage free way even with a laser with pulse duration in the nanosecond (ns) regime[21]. The a-Si serves as a buffer layer for the laser irradiation allowing for indirect ablation of the covering SiO₂ at laser energies five times lower than that needed to make ablation of crystalline silicon[21]. At the same time, the residual double layer of a-Si/SiO₂ between the laser spots has been shown to be excellent for surface passivation[14]. Herein, we measure and optimize the
contact resistance of metal contacts formed with DC magnetron sputtering of aluminum through the laser ablated holes. The contact deposition technique is chosen from equipment availability, and could equally well have been made by evaporation or screen printing through the holes in the PECVD stack. Through two-dimensional device simulations based on the finite element method, we optimize the metal to semiconductor contact area fraction with the obtained resistance values, and investigate the impact the contact resistance and area fraction have on the performance of a BC-BJ silicon solar cell.

2. EXPERIMENTS

4 inch, round, n-type, 1-3 Ω-cm, CZ wafers with a thickness of 300 μm were used in all the experiments. Prior to the processing, all wafers were cleaned in 5% HF, Piranha (1:4, H₂O₂: H₂SO₄), and HCl. An overview of the sample preparation steps is shown in Figure 1.

2.1. Emitter formation

A boron emitter of 50 ± 2 Ω/□ was formed by BBr₃ diffusion in a tube furnace. The resulting sheet resistance and the homogeneity of the emitter were measured with a four point probe and with a Semilab WTC-100 tool.

2.2. a-Si and SiO₂ deposition

Layers of a-Si of varying thicknesses were deposited on the wafers by PECVD. We used an Oxford Plasmalab 133 parallel plate system. The a-Si layers were deposited at 230 °C and 300 mTorr, with a flow of undiluted silane, and a RF input power of 15 W. The thickness was varied by changing the deposition time only. On top of the a-Si layer, we deposited a layer of SiO₂. The SiO₂ was deposited in the same chamber at a temperature of 300 °C, a chamber pressure of 500 mTorr, and an input power of 50 W. The thickness of the PECVD layers was measured by a variable angle spectroscopic ellipsometer. The resulting thickness of the SiO₂ was found to be about 400 nm, while the thickness of the a-Si layer varied between 40 nm and 300 nm for the different samples.
2.3. Laser ablation

After the thin film deposition, the round wafers were cut into four different samples of equal size. On each sample, a Nd:YVO₄, diode pumped, frequency doubled green laser with ns pulses was used to locally ablate the PECVD layers. The 1/e² diameter of the laser spot in focus was 40 µm. Each sample was exposed to a matrix of laser pulses with the same fluence. We refer herein to the fluence at the peak of the gaussian beam profile. The pulse fluence was varied from 0.1 J/cm² to 1.0 J/cm². In this energy range, the pulse duration varies close to linearly from 117 ns at 1.0 J/cm² to 150 ns at 0.1 J/cm². The density of laser spots on the samples was 78.5 %, corresponding to a spot to spot distance equal to the 1/e² diameter, and thus a maximum circular area coverage without spot overlap. After the laser ablation, the samples were cleaned with a similar cleaning sequence as before the processing, but this time with a weaker HF solution to protect the SiO₂. The diameter of the ablated holes, which varies with the fluence and the thickness of the a-Si layer, was measured by visual inspection in an optical microscope. For laser fluences just above the ablation limit, the diameter of the opening is smaller than the spot diameter since only the central region of the laser beam has a fluence above the ablation limit.
2.4. Contact formation

The contacts were made by DC magnetron sputtering of aluminum in an argon plasma with a power of 3 kW and a chamber pressure of $3 \times 10^{-3}$ mbar. The sputtering was done through a shadow mask that was held mechanically in place over the sample to obtain the desired pattern of contact fingers. A sketch of the shadow mask is shown in Figure 2. The resulting finger width was $115 \pm 2 \mu m$. The uncertainty results mainly from the shadow mask itself and not from the process. On each sample, four rows of contact fingers of equal width, but with different spacing were deposited for contact resistance measurements. After sputtering, the samples were cut with laser into four rows of contact fingers with contacting pads on each side, see Figure 2. This was done to isolate the structures and avoid shunting.

FIG. 2: The shadow mask used for aluminum sputtering. The shadow mask was held mechanically in place to ensure good contact between the mask and the sample during sputtering. The figure is not to scale, neither does the number of thin fingers per line match the actual number on the shadow masks.
2.5. Measuring the contact resistance

The transfer length method was used to measure the contact resistance of the contacts through the laser ablated holes[22, 23]. The so-called transfer length, $L_T$, is a measure of the distance over which the current is transferred from the emitter into the contact and vice versa, see Figure 3. The relation between $L_T$ and the specific contact resistance, $\rho_c$, is given by;

$$L_T = \sqrt{\frac{\rho_c}{R_S}},$$

(1)

where $R_S$ is the emitter sheet resistance. This simple expression for $L_T$ requires a negligible resistance in the metal, which is a reasonable assumption in our case with sputtered aluminum compared to a 50 $\Omega/$square emitter. Boron concentrations of $10^{19}$ cm$^{-3}$ to $10^{20}$ cm$^{-3}$ typically give contact resistances at the order of $10^{-6}$ $\Omega$·cm$^2$ with state of the art deposition techniques[24, 25]. This corresponds to a transfer length of 1-2 $\mu$m with standard solar cell emitters. Engelhart et al. demonstrated contact resistances in the order of $10^{-4}$ $\Omega$·cm$^2$ after picosecond laser ablation of thermal SiO$_2$ on phosphorus emitters on p-type wafers[17]. They also showed a corresponding contact resistance of $10^{-2}$ to $10^{-1}$ $\Omega$·cm$^2$ on phosphorus emitters after ablation with a ns laser.

![Diagram of contact resistance measurement](image)

FIG. 3: Schematic drawing of the current, $I_0$, flowing into and out of a contact. $L_T$ is the so-called transfer length, which is a measure of the distance over which the current is transferred from the emitter into the contact pad and vice versa. $2W$ is the width of a contact bar, while $2W'$ is the distance between two contact bars.

The idea of the method is to measure the voltage drop between the contact pads on each side of the finger pattern when a constant current is sent through the setup from pad to pad, see Figure 2. If the emitter sheet resistance is known, the transfer length, and thus $\rho_c$,
can easily be calculated from the following equation [22]:

$$V_{AB} = 2nI_0R_S \left( \frac{L_T}{Z} \tanh \left( \frac{W/L_T}{W'} \right) + \frac{W'}{Z} \right),$$

(2)

where $V_{AB}$ is the voltage drop between the two contact pads on each end of a row. $n$ is the number of contact fingers (+1) on the same row, and $I_0$ is the current flowing through the contact pattern. $Z$ is the length of the contact finger as indicated in Figure 2. $W$ is half the width of a contact finger, while $W'$ is half the width of the area between the contact fingers, see Figure 3. In Equation 2, the metal fingers are assumed to be in direct contact with the underlying substrate over the whole contact area. Herein, the metal fingers are in direct contact with the boron emitter only in the laser ablated holes. Between the holes, the a-Si/SiO$_2$ stack isolates the aluminum from the silicon. We are therefore in need of a slightly modified version of Equation 2. In calculations, we have approximated the circular contact spots by squares with the same area. The reduced metal to semiconductor area is accounted for by introducing the metal coverage fraction, $f$. In addition, we need to include the increased emitter resistance due to the non-contacted area under the metal fingers. Our modified expression that takes these considerations into account then becomes:

$$V_{AB} = 2nI_0R_S \left( \frac{L_T}{Z} \tanh \left( \frac{W/\sqrt{f}}{L_T} \right) + \frac{W'}{Z} + \frac{a \cdot d'}{Z} \left( 1 + e^{-d a/L_T} \right) \right),$$

(3)

d is half the width of the (square) point contact, while $d'$ is half the width of the gap between two point contacts, see Figure 4. The last term, $\frac{a \cdot d'}{Z} \left( 1 + e^{-d a/L_T} \right)$, accounts for the additional emitter resistance. The factor $a$ adjusts for the small offset between three spot to spot distances ($3 \times 40 \, \mu m = 120 \, \mu m$) versus the width of the contact finger ($115 \, \mu m$).

3. EXPERIMENTAL RESULTS AND DISCUSSION

Our best measured specific contact resistance was found to be 0.8 m$\Omega$·cm$^2$, and a wide range of samples with different a-Si thicknesses ablated with different laser fluences were found to give specific contact resistances below 10 m$\Omega$·cm$^2$. The results are calculated as the average of the four different rows of contacts on each sample. The results are shown in Figure 5 as specific contact resistance as a function of fluence for different a-Si layer thicknesses. In two recent papers, we investigated the laser fluence ablation threshold as a function of a-Si thickness both experimentally [21], and theoretically [26]. The results displayed in Figure
FIG. 4: Schematic drawing of the actual (left) and assumed (right) pattern of contact pads. The area of the circular and quadratic contact pads are equal. Also indicated in the figure are the width of a square contact pad and the distance between the contact pads, $2d$ and $2d'$, respectively.

5 are in good agreement with the results of both papers. For laser fluences lower than the ablation limit for the different a-Si thicknesses, there is a very large specific contact resistance. This is indicated as $1 \Omega \cdot \text{cm}^2$ in Figure 5.

FIG. 5: The measured specific contact resistance as a function of laser fluence for different thicknesses of the a-Si layer. The thicknesses are given in nm. Samples irradiated with a laser fluence below the ablation limit, have been assigned a specific contact resistance of $1 \Omega \cdot \text{cm}^2$ in the figure.

The best result, 0.8 m$\Omega \cdot \text{cm}^2$, was obtained for an a-Si layer with a thickness of 180 nm and a laser fluence of 0.59 J/cm$^2$.

4. SIMULATIONS

To investigate how this optimized contact resistance influences the total performance of a BC-BJ silicon solar cell, we performed two-dimensional device simulations. We used the
Atlas device simulator within the Silvaco simulation framework\cite{27} to simulate the active area performance of the solar cell. The evaluation of optical losses was done externally to Atlas with the program TracPro\cite{28}. This was done due to limited possibilities of performing ray tracing with the full solar spectrum on structured surfaces in Atlas. The simulated unit cell structure is shown in Figure 8. Before starting the optimization, the error associated with the mesh was eliminated through a sequence of simulations with increasingly finer meshes.

4.1. Ray tracing

Ray tracing was performed to account for surface reflection, escaped light, parasitic metal absorption, and free carrier absorption. The front surface was assumed perfectly covered with upright pyramids with a base width of 4 $\mu$m and an angle of 54.7$^\circ$ of the side walls to the surface orthogonal. The pyramids are introduced to lower the overall surface reflection, where the pyramids are typically fabricated by etching of the silicon surface in a weak alkaline solution. This is the industry standard for monocrystalline silicon solar cells. The pyramids were assumed covered with an anti-reflection coating of 10 nm SiO$_2$ close to the surface and 60 nm SiN$_x$ on top of the oxide. The thickness of the silicon wafer was set to 200 $\mu$m both in the ray tracing simulations and in the device simulations discussed later. On the backside, the original doping profiles (discussed in the next section) were replaced by 300 nm deep uniform box profiles of phosphorus and boron. This was done to simplify the simulations while at the same time ensuring that the overall integrated free carrier absorption (FCA) in the uniform profiles coincides with that of the actual doping profiles used in the device simulations. The FCA was calculated according to a parameterization by Martin Green\cite{29} commonly used for solar cells:

$$\alpha_{fc} = 2.6 \times 10^{-18} \times n \times \lambda^3 + 2.7 \times 10^{-18} \times p \times \lambda^2,$$

where $n$ and $p$ are the electron and hole concentrations, respectively, and $\lambda$ is the wavelength in $\mu$m. We chose to cut off the optical simulations at 1100 nm due to very long absorption lengths, and thus computing times, above this wavelength. However, the intensity of the AM 1.5 solar spectrum is relatively low between 1100-1150 nm, and the error introduced by this cut-off is expected to underestimate the short circuit current density with around
0.2 mA/cm². At the same time, ray tracing has been reported to overestimate the antireflection effect of the pyramids[30]. The overestimation originates from the fact that light is not able to resolve the geometrical sharp corners of the textures, thereby allowing some direct transmission or reflection, not found in the ray tracing[30]. The main results from the ray tracing are shown in Figure 6. Primary reflection is the limiting optical loss mechanism at short wavelengths, while escaped light is dominating at long wavelengths. The absorbed light curve has been weighted against the standard AM 1.5 solar spectrum to obtain a modified spectrum to be used as an input to the device simulator, see Figure 7.

![Absorption/Reflection vs Wavelength](image1)

**FIG. 6:** The fraction of light absorbed as a function of wavelength as calculated with ray tracing. The major optical loss mechanisms are also indicated in the figure.

![Modified Spectrum](image2)

**FIG. 7:** The modified solar spectrum used as input into the Atlas device simulations shown together with the original AM 1.5 solar spectrum. In the modified spectrum, optical losses such as primary reflection, escaped light, metallic parasitic absorption, and free carrier absorption have been accounted for.
4.2. Device simulation

The unit cell used in the simulation of the BC-BJ solar cells is shown in Figure 8. The width of the unit cell is one half of the so-called pitch which describes the distance between two equal polarity contacts. Herein, we used a unit cell width of 1100 μm, corresponding to a backside pitch of 2200 μm. The rather large size of the backside structures ensures that low-cost, high-throughput production methods can be used to produce the BC-BJ silicon solar cells[4, 31, 32]. We have used a base resistivity of 2 Ω·cm, based on an optimization performed by McIntosh et al.[33]. The thickness of the wafer was, as in the ray tracing, set to 200 μm. The emitter coverage fraction on the back was set to 73 %, while the backside doping profiles were read in from files obtained by SIMS measurements of near optimized experimental diffusion profiles of 50 Ω/□. A front surface field with a peak doping concentration of $3.8 \times 10^{18}$ cm$^{-3}$, a Gaussian shape, and a junction depth of 1.44 μm was specified according to the results of Granek and Hermle[34–36]. The surface recombination velocity on the thermally oxidized front side was calculated according to Cuevas et al.[37] as;

$$S = S_0 \text{ for } N_D < N_{\text{ref}}$$
$$S = S_0 (N_D/N_{\text{ref}}) \text{ for } N_D \geq N_{\text{ref}},$$

(5)

where $S_0 = 70$ cm/s and $N_{\text{ref}} = 7 \times 10^{17}$ cm$^{-3}$, while $N_D$ is the phosphorus doping concentration. On the backside, we assume the non-metallized part to be passivated by a stack of PECVD amorphous silicon and silicon oxide. We have used our experimental results from [21] to calculate the surface recombination velocity as[7]

$$S_{\text{eff}} = \frac{J_{sc} N_A}{n_i^2}.$$ 

(6)

A built-in concentration dependent lifetime model in Atlas[38–40] was used to calculate the Shockley-Read-Hall recombination lifetime. The bulk lifetime was set to 1 ms for both polarity charge carriers. Apparent bandgap narrowing was accounted for by implementing another model of Cuevas et al.[16];

$$\Delta E_G = A \ln(N/N_{\text{ref}}).$$

(7)

$\Delta E_G$ is the change in bandgap due to the doping concentration, $N$. $A$ is a constant

$$A = 0 \text{ if } N < N_{\text{ref}}$$
$$A = 14 \text{ meV if } N \geq N_{\text{ref}},$$

(8)
and $N_{\text{ref}} = 1.4 \times 10^{17}\ \text{cm}^{-3}$. In addition, we used a standard model for Auger recombination[41] together with a concentration dependent mobility model[42].

FIG. 8: Schematic drawing of the unit cell used for two-dimensional device simulations in the Atlas device simulator within the Silvaco simulation framework. The optical losses were calculated externally to Atlas prior to the device simulation. FSF and BSF are the front surface and back surface fields, respectively. The drawing is simplified and not to scale.

5. SIMULATION RESULTS AND DISCUSSION

The simulated efficiency as a function of the metal coverage fraction is shown in Figure 9. The efficiency increases with the contacted area fraction up to approximately 5 % where it remains close to constant at about 20.4 % up to 10 % area fraction. The increased efficiency with increasing metal coverage fraction is mainly a result of an increased fill-factor with reduced series resistance. The fill-factor as a function of the metal coverage fraction is shown in Figure 10, left axis.

The current-voltage (I-V) curve of the best simulated BC-BJ solar cell is shown in Figure 11. An efficiency of 20.4 % was obtained with the use of our latest experimental data based on a-Si backside passivation and ns laser ablation of local contact openings. However, there is still room for improvement. The diffusion profiles, which are based on experimental data and read in from file, are not fully optimized, and can still be further improved. In Figure 12, the local ideality factor, $m$, is plotted as a function of applied bias for different metal coverage fractions. The m-V curve describes the rate of change in the I-V curve. A high $m$-value indicates a slowly changing I-V curve, while a low $m$ value indicates a high rate of
change in the I-V curve. Near the maximum power point, the power loss in a silicon solar cell is usually dominated by series resistance, and a low $m$-value is desirable to ensure a sharp "knee-point" on the I-V curve. For a thorough introduction to m-V curves, see the thesis of K. McIntosh[43]. While the fill-factor is increasing with increasing metal coverage fraction, the open circuit voltage is reduced with increased contact area, mainly because of an increased total recombination current, see Figure 10, right axis. This opposes the increased fill-factor such that the efficiency remains more or less constant for a metal coverage fraction of 5-10%. For metallized area fractions above 10% (not shown in the Figure) the simulated efficiency slowly starts to decrease.

FIG. 9: The simulated efficiency as a function of the metal coverage fraction on the backside. The specific contacts resistance is set to 0.8 mΩ·cm², our best measured result.

FIG. 10: The fill-factor (left) and open-circuit voltage (right) as a function of the metal coverage fraction on the backside.
FIG. 11: The best cell result obtained from the two-dimensional device simulations. The cell reaches an efficiency of 20.4 %.

FIG. 12: The local ideality factor, $m$, as a function of applied bias for different metal coverage fractions. The increased ideality factor for low metal coverage at high voltage is a result of the series resistance.

6. CONCLUSION

We have shown that is possible to produce good quality contacts by laser ablation of a stack of a double layer of SiO$_2$ and a-Si and subsequent sputtering of aluminum to form local contacts through the ablated holes. The a-Si layer acts as a buffer layer for the laser irradiation, and makes it possible to make good quality contacts even with a laser with pulse duration within the nanosecond regime. The best result, 0.8 mΩ-cm$^2$, was obtained with an a-Si layer thickness of 180 nm and a laser fluence of 0.59 J/cm$^2$. We also investigated the impact the specific contact resistance would have on the performance of a back-contact back-junction silicon solar cell. The best simulated cell efficiency was found to be 20.4 % at a metal area coverage fraction of 5-10 %. However, we believe that there is still room
for improvement for this design. The diffusion profiles are not finally optimized, and the effective lifetime used in the simulations is somewhat modest.

7. ACKNOWLEDGEMENTS

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Paper VI

Tunneling in back-junction silicon solar cells

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Abstract

We have investigated tunneling mechanisms in back-junction silicon solar cells with adjacent highly doped regions of opposite polarity on the backside. This was exemplified by a back-contact back-junction silicon solar cell design, and performed through two-dimensional device and process simulations based on the finite element method. The effect of using these cells in a solar module was also investigated through SPICE simulations. Trap-assisted tunneling is found to have a negative influence on the performance of a silicon solar cell through a shunting-like effect, and thus resulting in a reduced fill factor. Band-to-band tunneling, on the other hand, leads to current breakdowns at low to moderate voltages, but does not influence the cell performance near the preferred operating point. This implies the heat distribution under reverse bias will be distributed throughout the whole junction area and that local hotspots can be avoided. These cells can thus be used in a solar module without the need for so-called bypass diodes. By careful optimization and tailoring of the doping profiles, it is possible to avoid any significant trap-assisted tunneling, while at the same taking advantage of band-to-band tunneling.
1. INTRODUCTION

Back-contacted back-junction (BC-BJ) silicon solar cells have both the emitter and the complete metallization on the rear side. This eliminates the shading present in conventional solar cells, and the sunward surface and back-side of the solar cell can therefore be independently optimized for optical and electrical performance, respectively. Moreover, having both contacts of opposite polarity on the same side may significantly simplify module assembly, and allow for an increased packing density of the cells\[1–3\]. A schematic drawing of a BC-BJ silicon solar cell can be seen in Figure 1. One of the main challenges in the processing of

![Schematic drawing of a back-contacted back-junction n-type silicon solar cell](image)

FIG. 1: Sketch of a back-contacted back-junction n-type silicon solar cell, The cell is shown upside-down, and is not to scale. BSF and FSF are the back- and front-surface field, respectively.

BC-BJ interdigitated solar cells is the alignment of the highly doped interdigitated p- and n-fingers on the back of the cell. Processing of early versions of BC-BJ silicon solar cells required four to six photolithography steps\[4\]. Photolithography is not regarded as compatible with large-scale processing of cost-competitive silicon solar cells, and the main interest in this type of cell design was for niche and academic reasons\[5\]. More recently, BC-BJ cell designs with self-aligned doping regions have been developed by both SunPower Inc. and Institut für Solarenergieforschung Hameln (ISFH) with their A-300\[6\], and RISE\[7\] silicon solar cells, respectively. Both cell designs consist of adjacent backside doping profiles of opposite polarity, and thus a potential lateral tunneling diode, depending on the doping profile and concentration\[8\]. Lately, another back-junction silicon solar cell concept was introduced by Harder et al.; the buried-emitter back-junction silicon solar cell\[9–11\]. This cell design has a back-junction covering almost the full cell area, and the simulations performed herein
are equally valid for the buried-emitter solar cell, though exemplified by a BC-BJ silicon solar cell design.

2. QUANTUM MECHANICAL TUNNELING

In a sufficiently strong electric field, charge carriers can tunnel through an energy barrier that it classically could not surmount. The effect is therefore often referred to as quantum mechanical tunneling. Under heavy doping and reverse bias, the band bending in a p-n junction will often be strong enough so that the valence band on the p-side of the junction overlaps with the conduction band on the n-side, see Figure 2. Tunneling of charge carriers can then take place inter-band, often referred to as band-to-band tunneling (BBT)[12]. Under very heavy doping, BBT is also known to dominate in light to moderately forward biased junctions, leading to a negative differential resistance region on the current-voltage (I-V) curve characteristic to BBT. Tunneling can also happen via traps in the bandgap, so-called trap-assisted tunneling (TAT). This effect can be important both under reverse and forward bias[13], including the range of the typically preferred operating point of a silicon solar cell. TAT can thus have a negative influence on the performance of a silicon solar cell through a shunting-like effect leading to a reduced fill factor. Implementation of the different tunneling mechanisms into the semiconductor transport equations is commonly done in two different ways. Either by a local model that simulates the tunneling with additional generation and recombination terms on either side of the junction to account for the changed carrier statistics, or by a non-local model that introduce an actual real spatial carrier transport through the barrier. A schematic drawing indicating the different tunneling mechanisms and the possible implementation is shown in Figure 2. In this paper we simulate a BC-BJ silicon solar cell with an adjacent interdigitated finger pattern through device and process simulations, and investigate the effect of quantum mechanical tunneling on the performance of a solar cell given different doping concentrations and profiles. The device and process simulations are based on the finite element method and performed with the ATLAS device simulator and the ATHENA process simulator, respectively, within the Silvaco simulation framework[14]. Further, we also investigate the performance of several of these cells put together in a solar module under different shading conditions.
FIG. 2: Sketch of an energy-band diagram indicating different tunneling mechanisms and the difference between local and non-local tunneling models.

3. DEVICE SIMULATIONS

To investigate the performance of a BC-BJ silicon solar cell we used the ATLAS device simulator. Tunneling of charge carriers usually takes place within a length in the order of 10 nm on each side of the junction, and a very dense mesh is therefore required in the tunneling region to be able to account for this effect. In ATLAS this is done by superimposing a very fine grid specifically for quantum mechanical effects on top an already defined dense mesh for standard simulations. To obtain results with a satisfactory accuracy, we isolated the simulations of the area around the tunneling junction shown in Figure 3. The tunneling current was deducted from simulated $I$-$V$ curves of this small junction with and without different tunneling mechanisms enabled. Further, the tunneling current was added to the total current of the full simulated unit cell of a BC-BJ (see Figure 3) silicon solar cell. The full unit cell includes the same junction area, but the density of the mesh is not sufficient to calculate an appropriated tunneling current through the junction. A small error is potentially introduced when adding the tunneling current density to the full device current density. However since the lateral p-n junction is the same in the two cases, the doping concentration near the junction is the important parameter, and the light-generated current only introduces a negligible error. The tunneling was assumed one-dimensional (lateral), and the spacing between the nodes in tunneling (x) direction was set to 0.3 nm in the background mesh and 0.03 nm
in the superimposed grid, ensuring smooth interpolations and stable convergent simulations.

For TAT simulations we used a local model thoroughly described by Hurks et al.[13]. The model is treated as conventional Shockely-Read-Hall recombination, but with an modified effective mass; $m^* = 0.25 \, m_0$ to account for the changed carrier statistics on each side of the junction.

![Diagram of a two-dimensional unit cell used for device simulations.](image)

**FIG. 3:** The two-dimensional unit cell used for the device simulations. The potential lateral tunneling junction between the emitter and the back-surface field (BSF) is indicated in the figure. FSF is the front-surface field.

To accurately model BBT Hermle et al. argued that a non-local model was needed to get satisfactory results when simulating tunneling in multijunction solar cells[15]. Non-local models also take into account the spatial variation of the energy bands, and thus a spatial flow of carriers through the junction. Such a model is readily available in ATLAS, and in the following we briefly explain how the non-local BBT is implemented into the carrier continuity equations in the device simulator. The net tunneling current for electrons with longitudinal energy $E$ and transverse energy $E_T$ across the junction can be expressed as [16]

$$J(E) = \frac{q}{\pi \hbar} \int \int T(E) [f_l(E + E_T) + f_r(E + E_T)] \rho(E_T) dE dE_T,$$

where $T(E)$ is the probability of tunneling for an electron with longitudinal energy $E$.

$$f_{l,r} = \frac{1}{\left(1 + \exp\left[\frac{E + E_T - E_{F,l,r}}{kT}\right]\right)},$$

is the Fermi-Dirac function using the quasi Fermi-level on the left ($l$) and right ($r$) side of the junction, respectively. $\rho(E_T)$ is the two-dimensional density of states corresponding to
the two transverse wavevector components given as

$$\rho(E_T) = \frac{\sqrt{(m_e m_h)}}{2\pi \hbar^2},$$

where \(m_e\) and \(m_h\) are the electron and hole effective masses, respectively. It can thus be shown[16] that the tunneling current in a transverse energy interval \(E - \Delta E/2\) to \(E + \Delta E/2\) is given as

$$J(E) \Delta E = \frac{q \sqrt{m_e m_h}}{2\pi \hbar^2} T(E) \ln(1 + \exp[(E_{FL} - E)/KT]) \ln(1 + \exp[(E_{FL} - E_{E_{max}})/KT]),$$

where \(E_{max}\) is the upper integration limit, see [16]. In the region near the junction where tunneling is possible, ATLAS calculates the evanescent wavevector at each node as

$$k(x) = \frac{k_e k_h}{\sqrt{k_e^2 + k_h^2}},$$

where

$$k_e(x) = \frac{1}{i\hbar} \sqrt{2m_e(x)E - E_c(x)},$$

and

$$k_h(x) = \frac{1}{i\hbar} \sqrt{2m_e(x)E_v - E(x)}.$$

The probability of tunneling, \(T(E)\), is then calculated as

$$T(E) = \exp \left( -2 \int_{x_{start}}^{x_{end}} k(x) \, dx \right)$$

As a starting point, we simulated \(J-V\) (current-density - voltage) curves of the small isolated tunneling junction (indicated in Figure 3) under illumination, with a uniform doping concentration of \(5 \times 10^{19}\) cm\(^{-3}\) on both sides of an abrupt junction. The result can be seen in Figure 4, where different curves originates from simulations of the same, small, lateral p-n junction, but with different tunneling models enabled. It is clearly seen that BBT is the dominating tunneling mechanism at reverse and small forward biases while TAT is dominating for larger forward biases. The characteristic negative differential resistance region in the \(J-V\) curve, which is found at small forward biases, is caused by BBT tunneling. Since TAT is dominating near and around the maximum power point, it will have a negative influence on the solar cell performance through a shunting-like behavior. A reduced shunt resistance will degrade the performance of a solar cell through a reduced fill factor. In the case of the
$J$-$V$-curves seen in Figure 4, the shunting will be severe, and the performance of a solar cell including this junction will be significantly degraded. The effect of decreasing the doping concentration to $1 \times 10^{19}$ cm$^{-3}$ in a similar p-n diode as described above is shown in Figure 5. BBT tunneling is no longer dominating at small forward and reverse biases. A larger reverse bias is required to obtain current breakdown, and the negative differential resistance region is no longer visible. TAT will, however, still dominate the $J$-$V$ curve in most of the forward bias region, and a solar cell including this junction will be completely shunted.

![Current-density–voltage curves of an illuminated tunneling junction](image)

FIG. 4: Current-density–voltage curves of an illuminated tunneling junction. The doping concentration was $5 \times 10^{19}$ cm$^{-3}$ on both sides of an abrupt junction. The different curves are results from simulations of the same p-n junction but with different tunneling mechanisms activated. BBT and TA are abbreviations for band-to-band and trap-assisted tunneling, respectively.

BBT tunneling does not influence the $J$-$V$ curves near the maximum power point (which is usually around 0.6 V for a high-efficiency silicon solar cell), and thus do not degrade the performance of a solar cell when operating near optimal conditions. BBT tunneling can actually have a positive effect on the performance of solar cell. Current breakdown at low or moderate reverse biases implies that local hot-spots can be avoided since the heat distribution under reverse bias will be distributed throughout the whole junction area. According to Hurkx et al. BBT was found to dominate the $J$-$V$ curve of p-n diodes under reverse biases down to a doping concentration of approximately $5 \times 10^{17}$ cm$^{-3}$, though under large reverse biases ($<-5$ V) for the lowest concentrations[13]. On the other hand, for doping levels higher than “a few times $1 \times 10^{18}$ cm$^{-3}$” they found TAT to significantly increase the non-ideal current through a diode. Thus, our tunneling junction should preferably have a doping concentration in the order of $1 \times 10^{18}$ cm$^{-3}$ to benefit from BBT at reverse bias.
FIG. 5: Current-density–voltage curves of an illuminated tunneling junction. The doping concentration was $1 \times 10^{19} \text{ cm}^{-3}$ on both sides of an abrupt junction. The different curves are results from simulations of the same p-n junction but with different tunneling mechanisms activated. BBT and TAT are abbreviations for band-to-band and trap-assisted tunneling, respectively.

while at the same time avoid undesired effects from TAT. In Figure 6 full solar cell $J$-V curves including both BBT and TAT as simulated with ATLAS is shown. The junction doping is indicated in the figure, and is assumed to be equal on both sides of an abrupt junction, see Figure 7. Reverse breakdowns at voltages $> -5V$ is starting to occur at a doping level of $2 \times 10^{18} \text{ cm}^{-3}$. In the simulation with a doping concentration of $3 \times 10^{18} \text{ cm}^{-3}$, TAT degrades the performance of the full BC-BJ silicon solar cell by approximately 0.1 % (absolute) compared to a cell without tunneling through the junction. It is worth
noting, however, that Shockley-Read-Hall recombination is most dominating when the two
different doping concentrations are equal, so this can be regarded as a "worst-case-scenario"
for TAT.

![Doping concentration profile](image)

**FIG. 7:** An abrupt doping profile used in the simulation of a BC-BJ silicon solar cell.

4. **MODULE SIMULATION**

In order to investigate how solar cells with a built-in tunneling junction behave when put
together into a module, a circuit simulation of a module containing 24 cells was performed
using SPICE. The $J-V$ curves obtained from the device simulation at different lighting
conditions, 1 sun, 1/2 sun and complete shading, for different sets of doping profiles were
implemented in the SPICE model using the *table* statement. A cell size of 156x156 mm$^2$
was assumed. We have investigated the case of optimal performance for each set of doping
profiles, as well as the cases where one or more cells are fully or partially shaded. The idea
is to use cells that do not show any significant TAT, while at the same time is dominated by
BBT tunneling at relatively moderate reverse biases. The corresponding heat dissipation will
be distributed throughout the whole junction area, and hot spots can be avoided without
the use of so-called bypass diodes. The cells used in the simulations had a peak doping
concentration of $5 \times 10^{19}$ cm$^{-3}$ while junction doping was $3 \times 10^{18}$ cm$^{-3}$ for both phosphorous
and boron. Several cases are shown in Figure 8 for one set of doping profiles. Here the full
module $I-V$ curves are shown for the cases with one, two and three partially shaded cells,
one completely shaded cell as well as for a non-shaded module. The $J-V$ curves for the
cell used in the module in Figure 8 with 1 sun, 1/2 sun and no irradiation is shown in
Figure 9. Comparing the case with one partly shaded cell in Figure 8 with the cell $J-V$
curve in 9 one may see that the reverse breakdown bias for one cell is roughly the difference in voltage between open-circuit voltage and the point where the module regains the full current. This would also be valid for a conventional module, however, in that case the reverse breakdown voltage would be larger and the cell reaching the breakdown voltage first would likely concentrate the full current through one or a few shunts resulting in irreversible damage to the cell or module from over-heating. The breakdown voltage in the case of a tunneling junction cell can be tuned to a certain extent. Full module current recovery will come at a lower voltage loss, hence a lower power loss, when reverse bias breakdown of the cells occur earlier. This is, however, an optimization problem as there is an increasing power loss in forward bias with a decreasing breakdown bias.

![Simulated I-V curve of a module with 24 156×156 mm² cells. I-V curves for the case of one and three partially shaded cells, one completely shaded cell as well as a non-shaded module is shown.](image)

FIG. 8: Simulated *I-V* curve of a module with 24 156×156 mm² cells. *I-V* curves for the case of one and three partially shaded cells, one completely shaded cell as well as a non-shaded module is shown.

As there are no local minima on the *I-V* curve, optimizing the power output of a module as described above should be quite similar to conventional modules.

5. **PROCESS SIMULATION**

The tunneling effect is very sensitive to the concentration profiles of the dopants. This underlines the importance of working with physically feasible profiles in the simulation
as the approximate analytical profiles may in this case yield results far away from reality. Therefore phosphorus and boron profiles were obtained from the process simulation software, ATHENA, within the Silvaco simulation framework. From the electrical simulation shown above, certain guidelines for a good tunneling junction have been concluded: the BBT needs high concentration of both dopants and sharp profiles, to minimize SRH recombination the concentrations at the junction should be an order of magnitude or so different, and to avoid TAT the dopant concentration right at the junction should be as low as possible. This has also been shown in the buried-emitter back-junction cell design of Harder et al.[9–11] to give a good emitter without significant loss due to TAT. A similar junction structure to that described in [11] was obtained in this work by first simulating a pre-deposition diffusion of boron at 950 °C for 15 min, with a subsequent drive-in at the same temperature for 20 min. Thereafter a phosphorus pre-deposition diffusion was simulated at 850 degrees C for 20 min. To obtain the sharp and buried junction with a low junction doping the boron deactivation effect of diffused hydrogen was taken into account. The hydrogen profile was approximated by a complementary error function profile with a depth parameter of 0.15 µm and a peak concentration of $10^{20}$ cm$^{-3}$ at the surface as a hydrogen diffusion model was not readily available in the ATHENA software. This corresponds roughly to the profile depth obtained by hydrogenation by a gas discharge plasma at 150 °C reported by Velichko et al.[17] and Johnsen et al.[18]. Further, it is assumed that the hydrogen present in the material will be completely bound by boron in a B-H complex except when there is an excess of H, thus passivating most of the boron close to the surface. The resulting active dopant concentration
profiles are shown in Figure 10.

![Graph showing active doping concentration of phosphorus and boron in a p-n junction.]

**FIG. 10:** The simulated active doping concentration of phosphorus and boron in a p-n junction as obtained with the ATHENA process simulator.

Finally, we tested this simulated junction profile to see how it influences the $J$-$V$ curve of a BC-BJ silicon solar cell. The curve is shown in Figure 11 and it shows similar features as the curves in Figure 6, namely negligible TAT but significant BBT at moderate reverse biases. The efficiency of this cell was degraded by 0.02 % absolute compared to the 18.56 % efficient reference cell. The small degradation was due to a very small reduction in the fill factor.

![Graph showing simulated current-density - voltage curve of a back-contact back-junction silicon solar cell.]

**FIG. 11:** Simulated current-density - voltage curve of a back-contact back-junction silicon solar cell including a lateral tunneling junction as simulated with the ATLAS device simulator. The junction profile used was obtained by process simulations with the ATHENA process simulator.
5.1. Conclusion

We have investigated quantum mechanical tunneling in back-junction silicon solar cells by device simulations with the Silvaco-ATLAS device simulator. We found that it is possible to keep unwanted trap-assisted tunneling at a negligible level while at the same time obtaining band-to-band tunneling effects at low to moderate reverse biases. This implies that the heat dissipation under reverse bias will be distributed throughout the whole junction area, and that local hot-spots can be avoided. We further demonstrated by SPICE module simulations how back-junction silicon solar cells have the possibility of operating in a solar module without the need for bypass diodes. We also showed how realistic doping profiles with the above mentioned advantages could be obtained by process simulations with the Silvaco-ATHENA process simulator. It was discussed that the junction doping concentrations should be in the order of $10^{18}$ cm$^{-3}$ to take advantage of these effects.