Industrial n-Type Silicon Solar Cells
with Co-Diffused Boron Emitters

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<tr>
<td>3D</td>
<td>three-dimensional</td>
</tr>
<tr>
<td>4PP</td>
<td>4-point probe</td>
</tr>
<tr>
<td>Ag/Al</td>
<td>aluminum containing silver</td>
</tr>
<tr>
<td>Al₂O₃</td>
<td>aluminum oxide</td>
</tr>
<tr>
<td>ALD</td>
<td>atomic layer deposition</td>
</tr>
<tr>
<td>APCVD</td>
<td>atmospheric pressure chemical vapor deposition</td>
</tr>
<tr>
<td>ARC</td>
<td>anti-reflection coating</td>
</tr>
<tr>
<td>B₂H₆</td>
<td>diborane</td>
</tr>
<tr>
<td>B₂O₃</td>
<td>boron trioxide</td>
</tr>
<tr>
<td>BBr₃</td>
<td>boron tribromide</td>
</tr>
<tr>
<td>BCl₃</td>
<td>boron trichloride</td>
</tr>
<tr>
<td>BDG</td>
<td>diethylenglycolmonobutylether</td>
</tr>
<tr>
<td>BO</td>
<td>boron-oxygen</td>
</tr>
<tr>
<td>BRL</td>
<td>boron-rich layer</td>
</tr>
<tr>
<td>BSF</td>
<td>back surface field</td>
</tr>
<tr>
<td>BSG</td>
<td>borosilicate glass</td>
</tr>
<tr>
<td>CB</td>
<td>conduction band</td>
</tr>
<tr>
<td>CO₂</td>
<td>carbon dioxide</td>
</tr>
<tr>
<td>CVD</td>
<td>chemical vapor deposition</td>
</tr>
<tr>
<td>Cz</td>
<td>Czochralski-grown</td>
</tr>
<tr>
<td>EB</td>
<td>etch-back</td>
</tr>
<tr>
<td>ECV</td>
<td>electrochemical capacitance-voltage measurement</td>
</tr>
<tr>
<td>EQE</td>
<td>external quantum efficiency</td>
</tr>
<tr>
<td>FCA</td>
<td>free carrier absorption</td>
</tr>
<tr>
<td>FELA</td>
<td>free energy loss analysis</td>
</tr>
<tr>
<td>FIB</td>
<td>focused ion beam</td>
</tr>
<tr>
<td>FSF</td>
<td>front surface field</td>
</tr>
<tr>
<td>FZ</td>
<td>float-zone</td>
</tr>
<tr>
<td>H₂O₂</td>
<td>hydrogen peroxide</td>
</tr>
<tr>
<td>Abbreviation</td>
<td>Definition</td>
</tr>
<tr>
<td>--------------</td>
<td>------------</td>
</tr>
<tr>
<td>H$_2$SO$_4$</td>
<td>sulfuric acid</td>
</tr>
<tr>
<td>HCl</td>
<td>hydrochloric acid</td>
</tr>
<tr>
<td>HF</td>
<td>hydrofluoric acid</td>
</tr>
<tr>
<td>HJT</td>
<td>heterojunction technology</td>
</tr>
<tr>
<td>HNO$_3$</td>
<td>nitric acid</td>
</tr>
<tr>
<td>IBC</td>
<td>interdigitated back contact cell</td>
</tr>
<tr>
<td>ICP</td>
<td>inductively coupled plasma</td>
</tr>
<tr>
<td>IQE</td>
<td>internal quantum efficiency</td>
</tr>
<tr>
<td>ITRPV</td>
<td>international technology roadmap for photovoltaics</td>
</tr>
<tr>
<td>IV</td>
<td>current-voltage</td>
</tr>
<tr>
<td>KOH</td>
<td>potassium hydroxide</td>
</tr>
<tr>
<td>LCO</td>
<td>local contact opening</td>
</tr>
<tr>
<td>LCoE</td>
<td>levelized cost of electricity</td>
</tr>
<tr>
<td>LID</td>
<td>light-induced degradation</td>
</tr>
<tr>
<td>LIT</td>
<td>lock-in thermography</td>
</tr>
<tr>
<td>MPP</td>
<td>maximum power point</td>
</tr>
<tr>
<td>N$_2$O</td>
<td>nitrous oxide</td>
</tr>
<tr>
<td>NaNO$_2$</td>
<td>sodium nitride</td>
</tr>
<tr>
<td>NH$_3$</td>
<td>ammonia</td>
</tr>
<tr>
<td>NO</td>
<td>nitrogen monoxides</td>
</tr>
<tr>
<td>PA</td>
<td>plasma-assisted</td>
</tr>
<tr>
<td>PECVD</td>
<td>plasma-enhanced chemical vapor deposition</td>
</tr>
<tr>
<td>PERC</td>
<td>passivated emitter and rear cell</td>
</tr>
<tr>
<td>PERT</td>
<td>passivated emitter rear totally diffused</td>
</tr>
<tr>
<td>POCl$_3$</td>
<td>phosphorous oxychloride</td>
</tr>
<tr>
<td>PSG</td>
<td>phosphor silicate glass</td>
</tr>
<tr>
<td>PV</td>
<td>photovoltaics</td>
</tr>
<tr>
<td>PVD</td>
<td>physical vapor deposition</td>
</tr>
<tr>
<td>QSS</td>
<td>quasi-steady-state</td>
</tr>
<tr>
<td>RCs</td>
<td>recombination centers</td>
</tr>
<tr>
<td>Abbreviations</td>
<td></td>
</tr>
<tr>
<td>---------------</td>
<td></td>
</tr>
<tr>
<td>RJ</td>
<td>rear junction</td>
</tr>
<tr>
<td>SCR</td>
<td>space charge region</td>
</tr>
<tr>
<td>SEM</td>
<td>scanning electron microscopy</td>
</tr>
<tr>
<td>s-FSF</td>
<td>selective front surface field</td>
</tr>
<tr>
<td>Si</td>
<td>silicon</td>
</tr>
<tr>
<td>SiNₓ:H</td>
<td>hydrogen-rich silicon nitride</td>
</tr>
<tr>
<td>SiO₂</td>
<td>silicon dioxide</td>
</tr>
<tr>
<td>SiOₓ:B</td>
<td>boron-doped silicon oxide</td>
</tr>
<tr>
<td>SiH₄</td>
<td>silane</td>
</tr>
<tr>
<td>SRH</td>
<td>Shockley-Read-Hall</td>
</tr>
<tr>
<td>STC</td>
<td>standard testing conditions</td>
</tr>
<tr>
<td>TLIM</td>
<td>two-light intensity method</td>
</tr>
<tr>
<td>TLM</td>
<td>transfer length method</td>
</tr>
<tr>
<td>VB</td>
<td>valence band</td>
</tr>
</tbody>
</table>
Symbols

- $a$: width/length of the unit cell [µm]
- $A$: area of the solar cell [cm$^2$]
- $A_c$: physical contact area [cm$^2$]
- $\Delta n/\Delta p$: excess charge carrier concentration of electrons/holes [cm$^{-3}$]
- $d_{ARC}$: thickness of anti-reflection coating [nm]
- $D_e/D_p$: diffusivity of electrons/holes [cm$^2$/s]
- $d_e$: emitter depth [µm]
- $d_F$: distance between fingers [mm]
- $d_S$: diameter of the contact spots [µm]
- $E_F$: Fermi energy [eV]
- $E_g$: bandgap energy of the semiconductor [eV]
- $E_t$: energy level of defects/traps within the bandgap [eV]
- $E_v$: photon energy [eV]
- $FF$: fill factor [%]
- $FF_0$: optimal fill factor [%]
- $f_c$: metallization coverage
- $f_{lh}$: fraction of low/highly doped area
- $f_{mask}$: fraction of the masked area
- $G$: generation rate of charge carriers [cm$^3$s$^{-1}$]
- $\eta$: energy conversion efficiency [%]
- $h_F$: height of the metal fingers [µm]
- $iV_{oc}$: implied open-circuit voltage [mV]
- $j_0$: dark saturation current density in one-diode model [fA/cm$^2$]
- $j_{01}$: saturation current density of first diode in two-diode model [fA/cm$^2$]
- $j_{01c}$: saturation current density of first diode below metal contacts [fA/cm$^2$]
- $j_{01lh}$: saturation current density of first diode in low/highly doped regions [fA/cm$^2$]
- $j_{02}$: saturation current density of second diode in two-diode model [nA/cm$^2$]
- $j_{02c}$: saturation current density of second diode below metal contacts [fA/cm$^2$]
- $j_{02lh}$: saturation current density of second diode in low/high doped regions [fA/cm$^2$]
<table>
<thead>
<tr>
<th>Symbols</th>
<th>Description</th>
</tr>
</thead>
<tbody>
<tr>
<td>$j_{0b}$</td>
<td>saturation current density of the base [fA/cm$^2$]</td>
</tr>
<tr>
<td>$j_{0e}$</td>
<td>saturation current density of the emitter [fA/cm$^2$]</td>
</tr>
<tr>
<td>$j_{0FSF}/j_{0BSF}$</td>
<td>saturation current density of the FSF/BSF [fA/cm$^2$]</td>
</tr>
<tr>
<td>$j_{0pass}$</td>
<td>emitter saturation current density of passivated areas [fA/cm$^2$]</td>
</tr>
<tr>
<td>$j_{0c}$</td>
<td>saturation current density below metal contacts [fA/cm$^2$]</td>
</tr>
<tr>
<td>$j_{MPP}$</td>
<td>current density at maximum power point [mA/cm$^2$]</td>
</tr>
<tr>
<td>$j_{ph}$</td>
<td>photo-generated current density [mA/cm$^2$]</td>
</tr>
<tr>
<td>$j_{sc}$</td>
<td>short-circuit current density [mA/cm$^2$]</td>
</tr>
<tr>
<td>$k_B$</td>
<td>Boltzmann constant [8.6173x10$^{-5}$ eV/K]</td>
</tr>
<tr>
<td>$\lambda_{in}$</td>
<td>wavelength of the incident light [nm]</td>
</tr>
<tr>
<td>$l_{BB}$</td>
<td>busbar length [mm]</td>
</tr>
<tr>
<td>$l_D$</td>
<td>distance between metallization and substrate edge [mm]</td>
</tr>
<tr>
<td>$l_F$</td>
<td>effective finger/contact length [mm]</td>
</tr>
<tr>
<td>$L_e/L_p$</td>
<td>mean diffusion length of electrons/holes [$\mu$m]</td>
</tr>
<tr>
<td>$l_S$</td>
<td>substrate length [mm]</td>
</tr>
<tr>
<td>$L_t$</td>
<td>characteristic length of the defect distribution [$\mu$m]</td>
</tr>
<tr>
<td>$L_T$</td>
<td>transfer length [$\mu$m]</td>
</tr>
<tr>
<td>$\mu_n/\mu_p$</td>
<td>mobilities of electron/holes [cm$^2$V$^{-1}$s$^{-1}$]</td>
</tr>
<tr>
<td>$n$</td>
<td>ideality factor in one-diode-model</td>
</tr>
<tr>
<td>$n_1$</td>
<td>ideality factor of first diode in two-diode-model</td>
</tr>
<tr>
<td>$n_2$</td>
<td>ideality factor of second diode in two-diode-model</td>
</tr>
<tr>
<td>$n_{ARC}$</td>
<td>refractive index of anti-reflection coating</td>
</tr>
<tr>
<td>$N_{BB}$</td>
<td>number of busbars</td>
</tr>
<tr>
<td>$N_D$</td>
<td>doping concentration [cm$^{-3}$]</td>
</tr>
<tr>
<td>$N_F$</td>
<td>number of fingers</td>
</tr>
<tr>
<td>$n_i$</td>
<td>intrinsic charge carrier concentration [cm$^{-3}$]</td>
</tr>
<tr>
<td>$N_S$</td>
<td>surface doping concentration [cm$^{-3}$]</td>
</tr>
<tr>
<td>$N_t$</td>
<td>density of defects [cm$^{-3}$]</td>
</tr>
<tr>
<td>$N_t^*$</td>
<td>effective density of defects [cm$^{-1}$]</td>
</tr>
<tr>
<td>$N_{i0}^*$</td>
<td>effective surface density of defects [cm$^{-1}$]</td>
</tr>
<tr>
<td>Symbol</td>
<td>Description</td>
</tr>
<tr>
<td>----------</td>
<td>-----------------------------------------------------------------------------</td>
</tr>
<tr>
<td>$p_{FF}$</td>
<td>pseudo fill factor [%]</td>
</tr>
<tr>
<td>$p_{in}$</td>
<td>incoming photon power density [W/cm$^2$]</td>
</tr>
<tr>
<td>$p_{MPP}$</td>
<td>power density at maximum power point [W/cm$^2$]</td>
</tr>
<tr>
<td>$q$</td>
<td>elementary charge [1.6022·10$^{-19}$ C]</td>
</tr>
<tr>
<td>$\rho_b$</td>
<td>base resistivity [Ωcm]</td>
</tr>
<tr>
<td>$\rho_c$</td>
<td>specific contact resistance [mΩcm$^2$]</td>
</tr>
<tr>
<td>$R$</td>
<td>recombination rate of charge carriers [cm$^{-3}$s$^{-1}$]</td>
</tr>
<tr>
<td>$R_{BB}$</td>
<td>resistance of the busbars [Ωcm$^2$]</td>
</tr>
<tr>
<td>$R_b$</td>
<td>resistance of the base [Ωcm$^2$]</td>
</tr>
<tr>
<td>$R_c$</td>
<td>contact resistance [Ωcm$^2$]</td>
</tr>
<tr>
<td>$R_D$</td>
<td>resistance of the doped layer [Ω]</td>
</tr>
<tr>
<td>$R_e$</td>
<td>lateral resistance of the emitter [Ωcm$^2$]</td>
</tr>
<tr>
<td>$R_{FSF}$</td>
<td>lateral resistance of the FSF [Ωcm$^2$]</td>
</tr>
<tr>
<td>$R_l$</td>
<td>line resistance of metal fingers [Ωcm$^2$]</td>
</tr>
<tr>
<td>$r_l$</td>
<td>line resistivity [mΩ/mm]</td>
</tr>
<tr>
<td>$R_M$</td>
<td>resistance of the TLM probes [Ω]</td>
</tr>
<tr>
<td>$R_P$</td>
<td>parallel/shunt resistance [Ωcm$^2$]</td>
</tr>
<tr>
<td>$R_S$</td>
<td>series resistance [Ωcm$^2$]</td>
</tr>
<tr>
<td>$R_{sh}$</td>
<td>sheet resistance [Ω]</td>
</tr>
<tr>
<td>$R_{sh,l}$</td>
<td>sheet resistance of the lowly doped area [Ω]</td>
</tr>
<tr>
<td>$R_{sh,h}$</td>
<td>sheet resistance of the highly doped area [Ω]</td>
</tr>
<tr>
<td>$R_T$</td>
<td>total resistance [Ω]</td>
</tr>
<tr>
<td>$\sigma_n/\sigma_p$</td>
<td>capture cross sections for electrons/holes [cm$^2$]</td>
</tr>
<tr>
<td>$S_n/S_p$</td>
<td>surface recombination velocity of electrons/holes [cm/s]</td>
</tr>
<tr>
<td>$\tau$</td>
<td>mean lifetime of the minority charge carriers [µs]</td>
</tr>
<tr>
<td>$T$</td>
<td>temperature [°C]</td>
</tr>
<tr>
<td>$T_S$</td>
<td>set peak firing temperature [°C]</td>
</tr>
<tr>
<td>$T_W$</td>
<td>wafer peak temperature [°C]</td>
</tr>
<tr>
<td>$V_{MPP}$</td>
<td>voltage at maximum power point [mV]</td>
</tr>
<tr>
<td>$V_{oc}$</td>
<td>open-circuit voltage [mV]</td>
</tr>
<tr>
<td>Symbols</td>
<td>Description</td>
</tr>
<tr>
<td>---------</td>
<td>-------------</td>
</tr>
<tr>
<td>$W$</td>
<td>wafer thickness [$\mu$m]</td>
</tr>
<tr>
<td>$w_{BB}$</td>
<td>busbar width [$\mu$m]</td>
</tr>
<tr>
<td>$w_{F}$</td>
<td>finger/contact width [$\mu$m]</td>
</tr>
<tr>
<td>$w_{M,F}$</td>
<td>finger width of etch-resist mask [$\mu$m]</td>
</tr>
<tr>
<td>$w_{M,BB}$</td>
<td>busbar width of etch-resist mask [$\mu$m]</td>
</tr>
</tbody>
</table>
1 Introduction

Since the conclusion of the world climate agreement in Paris in 2016 [1] at the latest, attended by almost 200 nations, it is acknowledged worldwide that the emission of greenhouse gases by conventional power generation using fossil fuels causes drastic changes to the global climate. For the first time in history, these nations are now bound to climate protection by this full agreement. The transition to environmentally friendly and sustainable power generation poses one of the most important challenges for future decades. Germany is one of the pioneers in this field as it has promoted the energy transition according to the Renewable Energy Law (EEG) since the year 2000. Photovoltaics (PV) shows up as a promising candidate for sustainable power generation because it uses solar irradiance as an almost infinite source of energy. The nominal peak power output of all PV plants installed in Germany was estimated at 40.8 GW at the end of 2016, equating to around 14% of the cumulative capacity installed worldwide [2]. In Germany, the generation of electricity by all PV plants amounted to about 7.4% of the total net demand in 2016. This is still a small but annually increasing contribution to the total generation of electricity comprising all renewable sources, which is around 37% [3]. The capability of PV for the future relies on its high potential for reducing the levelized costs of electricity (LCoE), which are currently in the range of 0.07 – 0.11 €/kWh [4]. To make PV competitive in the future it is crucial to further reduce the LCoE, which depend strongly on the location of the PV plants. According to the learning curve, LCoE are predicted to reach a level of 0.04 – 0.06 €/kWh in 2025 and 0.02 – 0.04 €/kWh in 2050 [4]. Thus, PV would be cheaper than conventional power generation techniques. The LCoE can be lowered either by enhancing the energy conversion efficiency of the solar cells for the same production costs, or by reducing the fabrication costs at the same level of energy conversion efficiency. This can only be achieved by continuous improvements of the solar cell manufacturing processes and development of new technologies.

The majority of the global fabrication of crystalline silicon (Si) solar cells is still based on p-type Si substrates as base material, mostly boron-doped [5]. However, boron-doped Si solar cells suffer from light-induced degradation (LID), which results in substantial power loss under illumination [6]. Furthermore, p-type Si substrates are sensitive to several of the common metal impurities which can accumulate during crystallization, such as iron, potentially limiting the minority charge carrier lifetime [7]. A way to overcome these drawbacks is to use n-type Si substrates as base material for the fabrication of crystalline Si solar cells. It is widely believed that n-type Si has a higher efficiency potential compared to p-type Si, thanks to its favorable electrical properties in terms of lower sensitivity to several common metal impurities like iron [7]. This leads to high minority charge carrier lifetimes and diffusion lengths that are hardly affected by LID. A high diffusion length of the minority charge carriers that exceeds the wafer thickness is essential for highly-efficient Si solar cell concepts, especially for rear junction (RJ) solar cell technologies like the ones used in this work.
Using n-type Si substrates as base material, the main issues are the formation of p-type emitters, commonly boron-doped, with high electrical quality, sufficient surface passivation and contact formation. Boron emitter formation is frequently realized by thermal diffusion of boron atoms from a boron tribromide (BBr$_3$) source [8]. A considerable drawback of this method is that for boron emitter and phosphorus front/back surface field (FSF/BSF) formation, at least two high temperature and masking steps are necessary. This is time- and cost-consuming and may also result in a considerable degradation of the minority charge carrier lifetime in the bulk, since high temperature treatments are known to be detrimental, especially on Czochralski-grown (Cz) Si with high oxygen content [9].

This work is mainly focused on overcoming the above-mentioned issue by the development of an alternative method for the formation of the boron emitters by means of co-diffusion from doping layers manufactured by plasma-enhanced chemical vapor deposition (PECVD). With this approach, only one single high temperature diffusion step is required for boron emitter and phosphorus FSF/BSF formation. This enables a simplified process sequence for time- and cost-efficient fabrication of n-type passivated emitter rear totally diffused (PERT) solar cells, aiming for low LCoE. For contact formation on boron emitters, screen-printing of aluminum containing silver (Ag/Al) pastes is commonly used. It is known that the application of Ag/Al pastes results in significant power losses, potentially compensating the benefits of using n-type Si substrates. To improve the energy conversion efficiency of n-type PERT solar cells, the contact formation mechanism by means of screen-printing of Ag/Al pastes has to be understood and the limiting recombination at the Ag/Al metal contacts has to be minimized. Therefore, another key goal of this work is a detailed investigation of contact formation via screen-printing of Ag/Al pastes and its influence on the characteristics of bifacial n-type PERT Si solar cells, in dependence of the parameters of contact firing, in order to minimize the metallization-induced power loss.

Another attractive approach to avoid detrimental Ag/Al metal contacts is the fabrication of RJ mono-facial n-type PERT solar cells. Since the boron emitter is formed on the rear side of these solar cells, an alternative metallization scheme can be utilized for contact formation. This is realized within this work by full-area deposition of Al after local contact opening (LCO) of the dielectric passivation layer stack via laser ablation. By this means, the metallization fraction can be reduced, compared to screen-printing, reducing the metallization-induced power loss due to recombination significantly. Otherwise, laser treatments on boron emitters may cause damage underneath the local contact spots and even in the space charge region (SCR) during LCO, inducing a considerable power loss due to enhanced recombination. This phenomenon is analyzed in this work on mono-facial n-type PERT solar cells and is validated by developing an empirical model for describing this damage, potentially induced by LCO and subsequent deposition of Al, and its influence on the characteristics of the solar cells.

This work is divided into three chapters. In the following second chapter, the structure, working principle and characteristics of crystalline Si solar cells as well as loss mechanisms and their influence on the characteristics of the solar cells are outlined. Focus is set on the development
of Si solar cells based on n-type Si substrates. Following that, techniques for boron emitter formation and the alternative approach for emitter formation by means of co-diffusion from doping layers fabricated by PECVD as well as the characterization of the electrical quality of these emitters are presented. In addition, the necessity of forming a phosphorus-doped FSF/BSF to improve the conversion efficiency of the n-type PERT solar cells is discussed. In the third chapter, bifacial n-type PERT Si solar cells with co-diffused boron emitters are analyzed. Focus is set on the metallization by means of screen-printing and the thereby induced power losses, targeting optimal front grid structure and contact firing conditions to minimize power losses induced by recombination and power losses due to resistance. A detailed loss analysis of the bifacial n-type PERT solar cells is presented. In chapter four, the investigations on n-type PERT solar cells are reviewed. Attention is turned on the LCO process using laser ablation and its compromising influence on the characteristics of the solar cells. For the first time, an empirical model for the distribution of defects underneath the contact spots, potentially generated during local contact formation, is presented. This model allows to successfully reproduce the measured characteristics of the n-type PERT solar cells by numerical simulation.
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In this chapter, the structure, working principle and current-voltage characteristics of crystalline silicon (Si) solar cells are outlined. Furthermore, optical, recombination and resistance related power loss mechanisms, which limit the energy conversion efficiency $\eta$ of the solar cells, are discussed in detail. Understanding and quantifying these power loss mechanisms is crucial for improving the solar cell fabrication processes and development of new technologies, allowing to reduce the levelized costs of electricity (LCoE). The focus is set on n-type crystalline Si, since it is used within this work as base material for the development of passivated emitter rear totally diffused (PERT) Si solar cells. An alternative approach for forming boron emitters by means of sequential *co-diffusion* from doping layers fabricated by plasma-enhanced chemical vapor deposition (PECVD) is presented, enabling considerable simplifications of the solar cell fabrication processes. Moreover, the characterization of the electrical quality of the *co-diffused* boron emitters and the formation of a phosphorus-doped front/back surface field (FSF/BSF), as well as methods for surface passivation are reviewed. Further information on Si solar cells can be obtained from reference literature on photovoltaics (PV) [10-14].

2.1 Functionality and structure

The conversion of light into electrical energy in crystalline solar cells is based on the generation of electron-hole pairs, known as internal photoelectric effect, and separation of the generated charge carriers. Electrons in the valence band (VB) can be excited into the conduction band (CB) by the absorption of incident photons with energy $E_\nu$, which has to be higher or equal to the bandgap energy $E_g$ of the semiconductor ($E_g \approx 1.11$ eV for Si at 300 K [13]). Photons with lower energy ($E_\nu < E_g$) will not be absorbed in the semiconductor and do not contribute to electrical power generation. If $E_\nu$ exceeds $E_g$, electrons in the VB will be excited into higher energy states in the CB and subsequently release their excessive energy by thermal relaxation. The mechanism of charge carrier generation by absorbing incident photons is depicted in Figure 2.1.

**Figure 2.1:** Generation of charge carriers by absorption of incident photons. Electrons in the valence band (VB) are excited into the conduction band (CB) if the photon energy $E_\nu$ is higher or equal to the bandgap energy $E_g$ of the semiconductor.
The photo-generated charge carriers are separated spatially by the electric field, induced by a doping gradient. Doping is realized by incorporation of foreign atoms into the Si crystal lattice, which leads to an excess concentration of electrons (majority charge carriers) and lack of holes (minority charge carriers) for n-type Si and reversed for p-type Si. At the transition from p- to n-type Si (pn-junction), the charge carrier concentration gradients lead to diffusion of electrons from the n- to the p-type region and diffusion of holes from the p- to the n-type region. The remaining fixed ionized doping atoms form the space charge region (SCR). Thus, an electric field is induced, leading to a field current which flows contrary to the carrier diffusion current. In thermal equilibrium, the field current equals the diffusion current. The Fermi energy $E_F$ has to be balanced in n- and p-type regions in thermal equilibrium, leading to bending of the energy bands at the SCR, as depicted in Figure 2.2. Under illumination conditions, individual Fermi levels (quasi-Fermi levels) must be considered [13]. The difference between the quasi-Fermi levels of electrons and holes corresponds to the open-circuit voltage $V_{oc}$ of the Si solar cell.

![Figure 2.2: Schematic band scheme of a p$^+$nn$^+$ structure used for n-type PERT Si solar cells with p$^+$-doped emitter and n$^+$-doped BSF.](image)

For the majority of industrially manufactured Si solar cells, p-type Si substrates are used as base material (usually boron-doped) [5]. However, the application of n-type Si substrates has several advantages compared to p-type Si substrates, as discussed in section 2.3.4. This work is focused on the development of n-type PERT (n-PERT) Si solar cells. On one side of the n-PERT solar cells, a high p-type doping ($p^+$) is used which forms the emitter, as shown in Figure 2.2. If the photo-generated minority charge carriers in the base (holes in the case of n-type Si) reach the SCR by thermal diffusion before recombining, they are forced by the electric field into the emitter, where they are majority charge carriers. The different mechanisms of charge carrier recombination are addressed in section 2.3.2. On the other side of the n-PERT solar cells, a highly n-type doped ($n^+$) region is formed which generates an electric FSF/BSF, hindering holes from reaching the wafer surface and thereby from recombining. The working principle and formation of a FSF/BSF are outlined in section 2.5.5. Because of charge carrier separation, $V_{oc}$ builds up between front and rear surface. Both sides are contacted by metal contacts (not shown in Figure 2.2), as discussed in section 3.3. If the solar cell is loaded by an external consumer, a
current $I$ flows through the metal contacts. The following section deals with the current-voltage (IV) characteristics of crystalline Si solar cells.

### 2.2 IV-characteristics

The characteristic IV-curve of an ideal Si solar cell without illumination, known as dark one-diode IV-curve, can be derived from the diffusion currents of a loaded pn-junction following an exponential law, first described by Shockley [15]

$$j(V) = j_0 \left( e^{\frac{qV}{n k T}} - 1 \right),$$  \hspace{1cm} (2.1)

with the Boltzmann constant $k_B$, elementary charge $q$, temperature $T$ in Kelvin and voltage drop $V$ across the cell. Usually, the current density $j = I/A$, which is independent of the solar cell area $A$, is used instead of $I$. The dark saturation current density $j_0$ is determined by the recombination current densities of emitter $j_{0e}$, base $j_{0b}$ and FSF/BSF $j_{0FSF/0BSF}$. Under illumination, the IV-curve shifts downwards by the photo-generated current density $j_{ph}$, as the photo-generated current flows in the opposite direction to the forward direction of the diode, as illustrated in Figure 2.3. Its value equals the short-circuit current density $j_{sc}$ at short-circuit conditions ($V = 0$)

$$j(V) = j_0 \left( e^{\frac{qV}{n k T}} - 1 \right) - j_{sc}. \hspace{1cm} (2.2)$$

The one-diode ideality factor $n$ can be calculated to $n = 1$ [15], when considering only radiative (band-to-band), Auger, Shockley-Read-Hall (SRH) and surface recombination. The different recombination mechanisms are discussed in section 2.3.2. In order to additionally account for recombination in the SCR, values of $n > 1$ have to be assumed. Solving equation (2.2) for $V$ at open-circuit conditions ($j = 0$) and neglecting the recombination in the SCR ($n = 1$), as well as resistance related power losses, leads to the open-circuit voltage

$$V_{oc} = \frac{k_B T}{q} \ln \left( \frac{j_{sc}}{j_0} + 1 \right) \approx \frac{k_B T}{q} \ln \left( \frac{j_{sc}}{j_0} \right). \hspace{1cm} (2.3)$$

This equation clarifies that an increase of the total $j_0$ of the Si solar cells, caused by the different recombination mechanisms, results in a decrease of $V_{oc}$. Therefore, quantifying and minimizing the $j_0$ values of the different areas of interest (emitter, base and FSF/BSF) is one of the most important issues of this work for improving the fabrication processes of the n-PERT solar cells, targeting high conversion efficiency and low LCoE. A widely used method for the experimental determination of the $j_0$ values is introduced in section 2.5.4.

The power output of a solar cell has its maximum value at the maximum power point (MPP) with the voltage $V_{MPP}$ and current density $j_{MPP}$. The maximal extracted power density $p_{MPP}$ can be defined as

$$p_{MPP} = j_{MPP} \cdot V_{MPP} = j_{sc} \cdot V_{oc} \cdot FF, \hspace{1cm} (2.4)$$
with the fill factor $FF$ of a solar cell given by

$$FF = \frac{j_{MPP} \cdot V_{MPP}}{j_{sc} \cdot V_{oc}}.$$  (2.5)

Figure 2.3 illustrates the correlation of the above-mentioned parameters. Note that because of general convention the values of the current density are illustrated with opposite sign.

Green published an empirical equation for the calculation of the optimal $FF$ ($FF_0$) of a solar cell, which is free of resistance related and recombination power losses, only by the knowledge of $V_{oc}$ of the solar cell [16]

$$FF_0 = \frac{qV_{oc}}{k_BT} - \ln \left( \frac{qV_{oc}}{k_BT} + 0.72 \right) \frac{qV_{oc}}{k_BT} + 1.$$  (2.6)

The $FF_0$ values are used for a fill factor loss analysis of bifacial n-PERT solar cells, presented in section 3.4.4, to clarify the origin of the $FF$ loss of the solar cells.

The most important parameter of a solar cell in describing its electrical quality is the energy conversion efficiency $\eta$, which can be calculated by the ratio of the maximum power density generated at MPP and the incoming photon power density $p_{in}$

$$\eta = \frac{p_{MPP}}{p_{in}} = \frac{j_{MPP} \cdot V_{MPP}}{p_{in}} = \frac{j_{sc} \cdot V_{oc} \cdot FF}{p_{in}}.$$  (2.7)

assuming $p_{in} = 0.1$ W/cm$^2$ for an intensity of 1 sun, $T = 25^\circ$C and AM1.5g spectrum as standard testing conditions (STC) [17].

![Figure 2.3: Dark and illuminated IV-curve of a crystalline Si solar cell. Note that because of general convention the values of the current density are illustrated with opposite sign.](image)

A real solar cell shows a non-ideal behavior of the pn-junction and additionally suffers from power losses induced by resistance. It has been shown in [18] that SRH recombination, induced
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by defect states located in the middle of the band gap in the SCR, can be described by a diode with an ideality factor $n = 2$, dominating the IV-curve of a real solar cell for low forward bias. Thus, a second diode with an ideality factor $n_2 = 2$ and a parasitic shunt resistance $R_p$ are added in parallel and a series resistance $R_S$ in line with the first diode with an ideality factor $n_1 = 1$. The equivalent circuit diagram is depicted in Figure 2.4. This leads to the two-diode model

$$j(V) = j_{01} \left( e^{\left( \frac{q(V-jR_S)}{n_1k_BT} \right)} - 1 \right) + j_{02} \left( e^{\left( \frac{q(V-jR_S)}{n_2k_BT} \right)} - 1 \right) + \frac{V-jR_S}{R_p} - j_{ph}, \quad (2.8)$$

describing the IV-characteristics of a real solar cell more precisely than the one-diode model in equation (2.2). The first term in equation (2.8) comprises all recombination mechanisms in base, emitter and FSF/BSF, also including the surface recombination of the solar cell, with the saturation current density of the first diode $j_{01}$. The second term describes the recombination in the SCR with the saturation current density of the second diode $j_{02}$. The third term includes the power loss caused by parasitic shunting of the pn-junction and thereby caused leakage currents, which also reduce the power output of the solar cell. Since the short-circuit current density $j_{sc}$ does not equal the photo-generated current density $j_{ph}$ anymore, because of resistance related power losses induced by $R_S, j_{ph}$ is used in equation (2.8) instead of $j_{sc}$ used in equation (2.2).

![Figure 2.4](image)

*Figure 2.4: Equivalent circuit diagram of the two-diode model, also considering recombination in the SCR as well as resistance related power losses for modeling a real solar cell.*

Fitting the two-diode model to experimentally determined IV-curves of the solar cells is used in this work for extracting the relevant solar cell characteristics, such as $j_{01}, j_{02}, R_S$ and $R_p$, thus enabling the quantification of the relevant loss mechanisms. This is required for understanding and minimizing these loss mechanisms by improving the solar cell fabrication processes, while targeting high $\eta$ of the n-PERT Si solar cells. Moreover, the whole characteristic IV-curve of a solar cell can be computed when knowing $j_{01}, j_{02}, R_S$ and $R_p$. This method is used in this work for the optimization of the front and rear side metallization grid of bifacial n-PERT solar cells presented in section 3.3.5.

The above-mentioned relevant loss mechanisms, limiting the solar cell parameters $j_{sc}, V_{oc}, FF$ and thereby $\eta$, are reviewed in detail in the following section.
2.3 Loss mechanisms

The energy conversion efficiency $\eta$ of a Si solar cell is directly correlated to its ability to absorb incident sunlight and generate electron-hole pairs, as described in section 2.1. All incoming photons that are reflected or transmitted do not contribute to power generation. These optical losses are outlined in the first subsection. The generated charge carriers have to be separated by the electric field of the SCR before recombining. The different recombination loss mechanisms are discussed in the second subsection. Furthermore, resistance related losses caused by $R_P$ and $R_S$ are analyzed in the third subsection. In order to develop solar cells with high $\eta$, all of these loss mechanisms must be minimized in general. The focus of this work is set on recombination and resistance related power loss mechanisms attributed to the metallization of n-PERT solar cells, since they still limit $\eta$ of these solar cells, especially for screen-printed bifacial devices.

2.3.1 Optical losses

All incident photons that do not generate electron-hole pairs account for optical losses. Optical losses mainly affect the short-circuit current density $j_{sc}$ and thereby $\eta$ of the solar cells. Four different optical loss mechanisms can be differentiated as discussed in the following:

I. Surface reflection:

Incident light can be reflected at the surface of the solar cell and does not contribute to the generation of electron-hole pairs and decreases $\eta$ of the solar cell. The plane surface of the Si substrate can be structured to prevent light from being reflected. Usually, a random pyramid texture is used for monocrystalline Si substrates. Thus, the angle of incidence is modified to give reflected light a second chance to strike the wafer surface again and to be absorbed. In addition, the surface reflectivity of the solar cells can be further reduced when using a dielectric anti-reflection coating (ARC). The ARC relies on the physical phenomenon of destructive interference of two light waves which are reflected at the two interfaces of the ARC. The ideal refractive index of the ARC $n_{ARC}$ can be estimated, according to the amplitude-condition with the refractive index of Si $n_{Si}$ and the refractive index of air $n_{air}$, to be

$$n_{ARC} = \sqrt{n_{air} \cdot n_{Si}}.$$  \hspace{1cm} (2.9)

For complete destructive interference of the incident light with wavelength $\lambda_{in}$, the ideal thickness of the ARC $d_{ARC}$ can be calculated, according to the phase-condition, by

$$d_{ARC} = \frac{\lambda_{in}}{4 \cdot n_{ARC}}.$$  \hspace{1cm} (2.10)

Commonly, a hydrogen rich silicon nitride (SiN$_x$:H) layer is used as ARC with $n_{ARC} \approx 2$ and $d_{ARC} \approx 75$ nm, because these parameters fit well to Si ($n_{Si} \approx 4$) and air ($n_{air} \approx 1$) for $\lambda_{in} \approx 600$ nm. This wavelength has been chosen since it is close to the maximal photon density of the spectral solar irradiance [11].
II. **Transmittance:**

If the rear side of the Si solar cell does not act as a perfect mirror, long-wave light can be transmitted through the solar cell, since Si is an indirect semiconductor with low absorption coefficient [11]. The transmitted light does not contribute to the generation of electron-hole pairs and decreases the solar cell efficiency. Since the wafer thickness has decreased in recent years due to Si material savings, transmittance losses seriously influence the performance of Si solar cells [19]. One way to deal with transmittance losses is applying a full-area rear side metallization, like full-area screen-printing of Al pastes, as used for the majority of the industrially manufactured p-type Si solar cells. However, the full-area metallization leads to high saturation current densities $j_0$ due to a high recombination rate at the rear surface. Transmittance losses can also be reduced by applying an optimized rear side passivation layer in terms of refractive index and thickness to maximize the rear side reflectivity. Light that is reflected at the rear side of the solar cell has a second chance to be absorbed. The morphology of the rear surface also affects the rear reflectivity. It is demonstrated in section 3.4.5 that the $j_{sc}$ loss due to transmittance is in the range of up to 0.5 mA/cm$^2$ for bifacial n-PERT Si solar cells.

III. **Shading:**

The front side metallization of solar cells causes optical power loss due to shading of the surface. Usually 4 - 8% of the surface of industrial solar cells is shaded by the front side metal contacts. The relative loss of $j_{sc}$ depends on the metallization fraction $f_c$ and can be lowered by reducing $f_c$. However, reducing $f_c$ usually results in an increase of the total series resistance $R_S$ of the solar cell, mainly attributed to an enhanced line and contact resistance of the metal fingers and/or to an increased lateral resistance between the metal fingers. Therefore, a compromise between power losses caused by shading and $R_S$ has to be made. This is further analyzed in section 3.3.

IV. **Parasitic absorption:**

Incoming light that is absorbed by parasitic absorption does not contribute to power generation in the solar cell. Although an ARC reduces the reflection of the incident light at the surface of the solar cell, the absorption in the ARC itself can also play an important role, if the imaginary part of the refractive index is high. For the passivation layers utilized in this work, such as aluminum oxide ($\text{Al}_2\text{O}_3$), thermal silicon dioxide ($\text{SiO}_2$) and SiN$_x$:H, parasitic absorption is negligible. Absorption in the metal contacts can also be described as parasitic absorption, but is commonly referred to as shading. Furthermore, free carrier absorption (FCA) must also be taken into account as parasitic absorption mechanism. Incident photons can, instead of generating electron-hole pairs, transmit their energy to electrons in the valence or conduction band which are excited to higher energy states. Note that FCA has no significant impact on conventional Si solar cells and also on the n-PERT solar cells studied in this work.
2.3.2 Recombination losses and minority charge carrier lifetime

Recombination can be described as the inverse process of the generation of electron-hole pairs. The recombination rate $R$ is proportional to the concentration of photo-generated excess charge carriers. Under illumination with generation rate $G$, the recombination rate $R$ determines the excess charge carrier concentration of electrons $\Delta n$ and holes $\Delta p$ (injection level)

$$\Delta p = G - R.$$  \hspace{1cm} (2.11)

The mean lifetime $\tau$ of the minority charge carriers (holes in n-type Si) can be defined by

$$\tau = \frac{\Delta p}{R}. \hspace{1cm} (2.12)$$

The mean diffusion length $L_p$ of minority charge carriers before recombination is given by

$$L_p = \sqrt{D_p \cdot \tau}, \hspace{1cm} (2.13)$$

with the diffusion coefficient $D_p$ of the minority charge carriers. In the following, the relevant recombination mechanisms are discussed:

I. **Shockley-Read-Hall (SRH) recombination:**

SRH recombination is caused by crystal defects (such as lattice vacancies, dislocations and grain boundaries), foreign atoms and precipitates. These defects induce admissible energy levels within the band-gap which act as recombination centers for electron-hole pairs. This recombination mechanism is known as *Shockley-Read-Hall* recombination, as Shockley, Read and Hall developed the mathematical description of this mechanism [20, 21]. The SRH recombination rate $R_{SRH}$ is given by

$$R_{SRH} = \frac{n \cdot p - n_i^2}{\tau_{p0} (n + n_1) + \tau_{n0} (p + p_1)}, \hspace{1cm} (2.14)$$

with the intrinsic charge carrier density $n_i$. The SRH time constants for the capture of electrons $\tau_{n0}$ and for the capture of holes $\tau_{p0}$ can be determined by

$$\tau_{n0} = \frac{1}{\sigma_n \cdot v_{th} \cdot N_t}, \hspace{1cm} \tau_{p0} = \frac{1}{\sigma_p \cdot v_{th} \cdot N_t}, \hspace{1cm} (2.15)$$

with the capture cross-sections for electrons $\sigma_n$ and holes $\sigma_p$, thermal velocity $v_{th}$ and defect density $N_t$. The SRH densities $n_1$ and $p_1$ can be calculated by the energy level $E_t$ within the bandgap induced by the defects

$$n_1 = n_i \cdot e^{\left(\frac{E_t - E_i}{k_B T}\right)}, \hspace{1cm} p_1 = n_i \cdot e^{\left(\frac{E_i - E_t}{k_B T}\right)}.$$  \hspace{1cm} (2.16)
continuum of defect energy states within the band gap, if the surface is not passivated. The surface recombination rate for a non-passivated surface is only limited by thermal diffusion of the minority charge carriers to the wafer surface. The density of dangling-bonds can be reduced by applying dielectric passivation layers (introduced in section 2.6), reducing the saturation current density $j_0$ and enhancing $\eta$ of the solar cells.

II. Radiative recombination:
It is also possible that an electron from the conduction band recombines directly with a hole from the valence band (band-to-band recombination), which is the inverse process to electron-hole pair generation. This recombination mechanism is also referred to as radiative recombination as the recombining electron releases its energy by emission of a photon (luminescence). The recombination rate for radiative recombination depends on the product of the charge carrier concentrations for electrons and holes and can be determined by [22]

$$R_{\text{rad}} = C_{\text{rad}}(n \cdot p - n_0 \cdot p_0), \quad (2.17)$$

with the temperature-dependent coefficient $C_{\text{rad}} = 9.5 \cdot 10^{-15}$ cm$^3$/s for $T = 300$ K [23]. This intrinsic recombination mechanism is unlikely for Si solar cells, as Si is an indirect semiconductor and a phonon is additionally required for this recombination process. Therefore, radiative recombination does not play a fundamental role for Si solar cells. The luminescence signal, which is generated by radiative recombination, can be utilized for the characterization of the spatial distribution of the minority charge carrier lifetime $\tau$. Since radiative recombination is a non-limiting recombination mechanism and the luminescence signal depends on the excess charge carrier concentration, according to equation (2.17), areas with a high luminescence signal can be qualified as regions with low total recombination rate, including all other recombination mechanisms.

III. Auger recombination:
The released energy from recombination of electron-hole pairs can also be transferred to electrons in the conduction band or holes in the valence band that get excited to a higher energy level. The excited charge carriers release their excess energy by thermal relaxation. For this recombination process, a second electron or hole becomes involved, which makes the recombination process less probable. The Auger recombination rate $R_{\text{Auger}}$ can be calculated by [24]

$$R_{\text{Auger}} = C_n(n^2 \cdot p - n_0^2 \cdot p_0) + C_p(n \cdot p^2 - n_0 \cdot p_0^2), \quad (2.18)$$

with the Auger-coefficients $C_n = 2.8 \cdot 10^{-31}$ cm$^6$/s and $C_p = 9.9 \cdot 10^{-32}$ cm$^6$/s [24]. Auger recombination can be neglected in the bulk of conventional Si solar cells for an injection level of around 1 sun illumination. Due to the high doping of emitter and FSF, Auger recombination must be considered in these regions.
Recombination in emitter, base and FSF/BSF can qualitatively be described by the saturation current density of the first diode $j_{01}$ in the two-diode model (see equation (2.8) in section 2.2). The saturation current density of an infinite n-type base $j_{0b}$ with doping concentration $N_D$ can be derived from the continuity equation to be [12]

$$
j_{0b} = \frac{q \cdot D_p \cdot n_i^2}{L_p \cdot N_D}.
$$

The saturation current density of the highly-doped emitter can be defined by the recombination current density $j_{surf}$ flowing to the wafer surface [25]

$$
j_{surf} = j_{0e} \left(\frac{n \cdot p - n_i^2}{n_i^2}\right) = j_{0e} \left(\frac{\Delta p (N_D + \Delta p) - n_i^2}{n_i^2}\right),
$$

with $j_{0e}$ including all of the recombination mechanisms in the emitter and at the wafer surface. Equation (2.20) can be simplified for high injection ($\Delta p \gg n_i$ and $\Delta p \gg N_D$) to

$$
j_{surf} = j_{0e} \left(\frac{\Delta p}{n_i}\right)^2.
$$

If the boundary surface between base and SCR is treated like an imaginary surface with the recombination rate $R_{surf}$ and effective surface recombination velocity $S_{eff}, j_{surf}$ can be calculated by

$$
j_{surf} = q \cdot R_{surf} = q \cdot \Delta p \cdot S_{eff},
$$

assuming negligible recombination in the SCR. Equating equation (2.20) and (2.22) results in

$$
j_{0e} = S_{eff} \left(\frac{q \cdot n_i^2}{N_D + \Delta p}\right).
$$

The sum of the saturation current densities of base $j_{0b}$, highly-doped emitter $j_{0e}$ and FSF/BSF $j_{0FSF/BSF}$ equals the total $j_{01}$ in the two-diode model.

The effective minority charge carrier lifetime $\tau_{eff}$ is determined by the recombination in the bulk and at the wafer surface

$$
\frac{1}{\tau_{eff}} = \frac{1}{\tau_{bulk}} + \frac{1}{\tau_{surf}}.
$$

For symmetrical wafers with the same surface recombination velocities $S$ at both wafer surfaces and wafer thickness $W$ follows [26]

$$
\tau_{surf} \approx \frac{W}{2S} + \frac{1}{D} \left(\frac{W}{\pi}\right)^2.
$$

The second term can be neglected for low surface recombination leading to

$$
\frac{1}{\tau_{eff}} = \frac{1}{\tau_{bulk}} + \frac{2S}{W}.
$$
For an emitter (or FSF/BSF) at the wafer surface, $S$ has to be replaced by $S_{\text{eff}}$ and equation (2.26) can be adapted with equation (2.23) to

$$
\frac{1}{\tau_{\text{eff}}} = \frac{1}{\tau_{\text{bulk}}} + \frac{2j_{0e}(N_D + \Delta p)}{q \cdot W \cdot n_{\text{i}}^2}.
$$

(2.27)

The first term can be neglected for very high $S$ values, whereby $\tau_{\text{surf}}$ is limited by the diffusion of charge carriers to the wafer surfaces in this case. Equation (2.27) is applied in this work for the determination of $j_{0e}$ to qualify the electrical quality of co-diffused boron emitters using the method of Kane and Swanson, as introduced in section 2.5.4.

### 2.3.3 Resistance related losses

As already mentioned in section 2.2, a real solar cell suffers also from power dissipation due to ohmic resistors. If the current $I$ flows through an ohmic resistance $R$, the power

$$
P = I^2 \cdot R
$$

(2.28)

is lost and transferred into heat. Thus, the maximal power output and therewith $\eta$ of the solar cell is decreased. Two kinds of resistance related power losses can be differentiated:

1. **Power loss due to series resistance:**

   According to the two-diode model discussed in section 2.2, the series resistance $R_S$ is connected in line with the two diodes. Therefore, the whole current has to flow through $R_S$. To minimize this power loss, $R_S$ must be kept as small as possible. In general, the transport of charge carriers in the Si substrate itself and the current flow through the metallization result in power losses due to heat dissipation. Figure 2.5 illustrates all components accounting for the total $R_S$ of a rear junction (RJ) n-PERT Si solar cell with phosphorus-doped FSF. The total $R_S$ of a RJ n-PERT solar cell reads

$$
R_S = R_{BB,f} + R_{l,f} + R_{c,f} + R_{FSF} + R_b + R_e + R_{c,b} + R_{lat,b},
$$

(2.29)

with the components

- $R_{BB,f}$: line resistance of the busbars (BB) on front side, if the BB are connected locally in the module or during IV-measurements,
- $R_{l,f}$: line resistance of the front contact metal fingers,
- $R_{c,f}$: contact resistance at the interface between metal fingers and FSF,
- $R_{FSF}$: lateral resistance of the FSF between metal fingers,
- $R_b$: resistance of the base substrate,
- $R_e$: lateral resistance of the emitter between local contacts,
- $R_{c,b}$: contact resistance at interface between emitter and rear side metallization,
- $R_{lat,b}$: lateral resistance of rear contacts, if the full-area rear contact is connected locally in the module or during IV-measurements.
Note that these values are in general area-normalized by multiplication with the total area $A$ of the solar cell, which allows reasonable comparison of these values for solar cells with different cell areas. For common H-grid metallization by screen-printing (see section 3.3) on the front side and full-area metallization on the rear side, line resistance of the BB and lateral resistance of the rear contact can be neglected. For bifacial n-PERT Si solar cells, the line resistance of the rear side contact metal fingers $R_{l,b}$ and BB $R_{BB,b}$ must also be implemented in equation (2.29) instead of $R_{lat,b}$. This is further discussed in section 3.4.5. Equation (2.29) shows that the metallization has an essential contribution to the total $R_S$ of n-PERT solar cells. Therefore, another issue of this work is to minimize power loss caused by heat dissipation attributed to the metallization of the solar cells. The different components of the metallization-induced $R_S$ and their impact on the total power loss of the solar cells are reviewed in section 3.3.

II. Power losses due to parallel resistance:
Parasitic leakage currents, also referred to as shunts, are mainly caused by insufficient edge isolation of the front and rear side of the solar cells and/or disruption of the pn-junction (e.g. penetration of the metallization through the SCR into the base or crystal defects such as grain boundaries). Parasitic shunts can be described, according to the two-diode model discussed in section 2.2, by a parallel respectively shunt resistance $R_P$ that is connected in parallel with the two diodes. If $R_P$ is high, the leakage currents are low. Therefore, the solar cell process steps must be optimized in order to achieve high $R_P$ values to minimize power losses. It is demonstrated in section 3.4 that the application of screen-printing aluminum containing silver (Ag/Al) pastes and subsequent contact firing causes shunting problems, which may lead to low $R_P$ values if the contact firing conditions are not optimized.

![Figure 2.5: Schematic illustration of all components contributing to the total series resistance $R_S$ of a mono-facial n-type PERT Si solar cell. In general, the transport of charge carriers in the Si substrate itself and the current flow through the metallization result in power losses due to heat dissipation.](image-url)
2.3.4 Influence on IV-characteristics

All power loss mechanisms discussed in the previous sections compromise the characteristics of Si solar cells. Since this work is focused on recombination and resistance related power losses attributed to the metallization, the impact of the most important parameters $R_S$, $R_P$, $j_{01}$ and $j_{02}$ on the solar cell characteristic IV-curve is analyzed in more detail in the following. The implicit formula in equation (2.8), describing the two-diode model, was solved with an iterative optimization algorithm known as generalized reduced gradient (GRG) method.

A photo-generated current density $j_{ph} = 39$ mA/cm$^2$ was used for all studies. To investigate the influence of only $R_S$ on the characteristic IV-curve of a solar cell, the current densities of the first and second diode were set to extremely low values ($j_{01} = 10$ fA/cm$^2$; $j_{02} = 1$ nA/cm$^2$) and the shunt resistance was set to $R_P = 1$ MΩ/cm$^2$. This leads to a negligible influence of $R_P$ and the recombination mechanisms on the IV-curve. The computed IV-curves for various $R_S$ values in the range of $0 – 50$ Ω/cm$^2$ as well as the solar cell characteristics $FF$ and $j_{sc}$, extracted from these IV-curves, are illustrated in Figure 2.6a. $FF$ of the solar cell is drastically reduced by increasing $R_S$ due to a shift of MPP towards lower $V_{MPP}$ caused by the voltage drop due to $R_S$. $V_{oc}$ is not affected by $R_S$ because no current is extracted from the solar cell under open-circuit conditions. For $R_S > 10$ Ω/cm$^2$, which leads to an ohmic behavior of the IV-curve, $j_{sc}$ is also decreased. Note that $R_S$ is usually in the range of $0.4 – 1.0$ Ω/cm$^2$ for industrially manufactured Si solar cells. Therefore, the main impact of $R_S$ on the solar cell is a limitation of $FF$.

For the study of the impact of $R_P$ on the IV-curve of a Si solar cell, the same assumptions for the current densities of the first and second diode were made again ($j_{01} = 10$ fA/cm$^2$; $j_{02} = 1$ nA/cm$^2$), while setting $R_S$ to zero in this case. Thus, $R_S$ has no and the recombination mechanisms only a negligible influence on the IV-curve of the solar cell. The computed IV-curves for various $R_P$ values in the range of $10 – 10000$ Ω/cm$^2$ and the solar cell characteristics $FF$ and $V_{oc}$, extracted from these IV-curves, are depicted in Figure 2.6b. $FF$ is significantly reduced by lowering $R_P$ due to a shift of MPP mainly towards lower $j_{MPP}$ caused by parasitic leakage currents. $j_{sc}$ is not affected, since all charge carriers are directly extracted at the metal contacts under short-circuit conditions. For $R_P < 100$ Ω/cm$^2$, $V_{oc}$ is also drastically lowered due to considerable parasitic leakage currents. Note that $R_P$ usually has no substantial influence on the IV-characteristics of industrially manufactured Si solar cells, since $R_P$ is commonly above $1$ kΩ/cm$^2$ for these solar cells.

Based on the characteristic impact of $R_S$ and $R_P$ on the IV-curve of Si solar cells, the fabrication processes have to be improved in order to minimize $R_S$ and maximize $R_P$. As mentioned in the previous subsection, a significant contribution to the total series resistance $R_S$ can be attributed to the metallization (see equation (2.29)). Therefore, the metallization processes by means of screen-printing of Ag and Ag/Al pastes is of key importance and analyzed in chapter 3. Due to the strong influence of $R_S$ on $FF$, the optimal front side metallization pattern is computed in section 3.3.5 in order to minimize metallization induced power loss of n-PERT Si solar cells.
To investigate the impact of $j_{01}$ on the characteristic IV-curve of a solar cell, the current density of the second diode was set to $j_{02} = 1 \text{nA/cm}^2$, assuming negligible recombination in the SCR. The ohmic resistors were set to $R_S = 0 \text{ } \Omega \text{cm}^2$ and $R_P = 1 \text{ M} \Omega \text{cm}^2$, leading to a negligible impact of the ohmic resistors on the IV-curve. The computed IV-curves for $j_{01}$ values in the range of $10^1 - 10^5 \text{ fA/cm}^2$ as well as the characteristics FF and $V_{oc}$, extracted from these IV-curves, are depicted in Figure 2.7a. It is evident that $j_{01}$ affects the IV-curve in the high voltage range. Thus, FF is only slightly affected by $j_{01}$, whereas $V_{oc}$ follows a distinct decay for increasing $j_{01}$ because of enhanced recombination in the emitter, base and FSF/BSF.

For the investigation of the influence of $j_{02}$ on the IV-curve of a solar cell, the current density of the first diode was set to $j_{01} = 10 \text{ fA/cm}^2$, assuming recombination mainly in the SCR. The ohmic resistors were again set to $R_S = 0 \text{ } \Omega \text{cm}^2$ and $R_P = 1 \text{ M} \Omega \text{cm}^2$. The computed IV-curves for $j_{02}$ values in the range of $10 - 10^5 \text{ nA/cm}^2$ as well as the values for FF and $V_{oc}$, extracted from these IV-curves, are depicted in Figure 2.7b. It is observed that FF and $V_{oc}$ are continuously...
decreasing for increasing \( j_{02} \) because of enhanced recombination in the SCR. \( j_{02} \) affects the IV-curve also in the low voltage range (200 – 500 mV). Therefore, its detrimental influence on \( FF \) and \( V_{oc} \) is even more pronounced for an increase in \( j_{02} \) compared to an increase in \( j_{01} \). However, typical values of \( j_{01} \) are in the range of 100 - 1000 fA/cm\(^2\) and values of \( j_{02} \) in the range of 10 – 100 nA/cm\(^2\) for industrially manufactured solar cells. Therefore, \( j_{01} \) is mainly limiting for \( V_{oc} \), while \( j_{02} \) primarily leads to a limitation of \( FF \) (compare absolute values of \( FF \) and \( V_{oc} \) in Figures 2.7a/b).

**Figure 2.7:** Influence of \( j_{01} \) and \( j_{02} \) on the characteristic IV-curve of a Si solar cell, assuming negligible influence of the ohmic resistors (\( R_S = 0 \) \( \Omega \)cm\(^2\) and \( R_P = 1 \) M\( \Omega \)cm\(^2\)). \( V_{oc} \) and \( FF \) are drastically reduced for high \( j_{01} \) and \( j_{02} \) values. The compromising influence on \( V_{oc} \) and \( FF \) is even more pronounced for an increase in \( j_{02} \) compared to an increase in \( j_{01} \). Note that due to general convention the values of the current density with opposite sign are illustrated. The lines serve as a guide to the eye.

Besides its strong influence on \( R_S \) and \( R_P \), the metallization has also an essential impact on the total \( j_{01} \) and \( j_{02} \) of the Si solar cells. Especially when using Ag/Al pastes, the metallization-induced recombination can be limiting for the performance of the Si solar cells. Therefore, the focus is set on contact formation using screen-printing of Ag/Al metal pastes for contacting boron emitters. A detailed analysis thereof is presented in section 3.4.
2.4 Solar cells with n-type base material

Usually, p-type Si substrates are used as base material for the fabrication of industrial crystalline Si solar cells with a market share of almost 95% in the year 2016 [5]. It is predicted by the international technology roadmap for photovoltaics (ITRPV) that the global market share of monocrystalline n-type Si substrates as base material increases to > 30% in the next decade [5].

The predominance of p-type Si substrates in the past can be explained by the excellent doping ability of boron, usually used for p-type doping, due to its high segregation coefficient in Si of 0.8, ensuring a homogeneous doping over the whole ingot [27]. Moreover, the negatively doped emitter can be formed by the well-developed and controllable phosphorus oxychloride (POCl₃) diffusion. A promising solar cell technology based on p-type Si is the passivated emitter and rear cell (PERC) technology, resulting in solar cell efficiencies of more than 22% for large area solar cells [28].

However, there are several disadvantages associated with boron-doped p-type Si substrates, such as light-induced degradation (LID), resulting in substantial power losses under illumination [6, 29-32]. The LID mechanism is based on the activation of boron-oxygen (BO) related recombination defects under illumination which drastically decrease the minority charge carrier lifetime. The concentration of the BO-related defects is linearly correlated to the boron concentration in uncompensated crystalline Si substrates [29]. It was demonstrated that the detrimental decrease of the minority charge carrier lifetime induced by LID can be permanently regenerated by additional process steps [33-38]. The regeneration process converts the recombination active BO-defects into inactive states. Current investigation shows that regeneration can be conducted in an in-line process within seconds making it industrially feasible [35, 36]. Another drawback of p-type Si substrates is the higher sensitivity to several of the common metal impurities that can be accumulated during crystallization such as iron which potentially limit the minority charge carrier lifetime [7].

To overcome these disadvantages, n-type Si substrates can be applied as base material for Si solar cell production. One reason for the increasing demand of n-type Si is its higher efficiency potential thanks to its favorable electrical properties compared to p-type Si [7]. In particular, the lower sensitivity of n-type Si to some of the common metal impurities like iron due to the lower capture cross section for holes σₚ, which are the minority charge carriers in n-type Si, and accordingly the lower SRH recombination rates (see equations (2.14 - (2.16)) at these defect sites [7] lead to higher minority charge carrier lifetimes. A low recombination rate in the base is mandatory for high-efficient RJ Si solar cell technologies like the one used in this work since high minority charge carrier lifetimes and diffusion lengths that exceed the wafer thickness are required. This is based on the fact that the charge carriers from the front surface, where most of the charge carriers are generated, must reach the rear side emitter, where the charge carriers are separated by thermal diffusion before recombining (see section 2.1). Another advantage of n-type Si is the weak influence or absence of LID because the concentration of the BO-related
defects is linearly correlated to the boron concentration in uncompensated Si substrates [39]. Therefore, it is negligible in uncompensated phosphorus-doped n-type Si substrates like the ones used in this work.

The majority of the industrial production of crystalline Si solar cells based on n-type Si base substrates has been limited so far to advanced technologies, such as the interdigitated back contact cell (IBC) concept, reaching conversion efficiencies of up to 25% [40], and combined with heterojunction technology (HJT), enabling conversion efficiencies up to 26.6% [41]. This confirms the high potential of n-type crystalline Si substrates. However, these advanced solar cell concepts require complex process sequences. A promising approach for reducing the LCoE is the combination of high potential n-type Si substrates with low-cost industrial manufacturing technologies used for many p-type solar cell concepts, such as screen-printing and POCl₃ gas-phase diffusion. The POCl₃ diffusion can be utilized to form a phosphorus-doped FSF/BSF, which is mandatory for minimizing the front/rear surface recombination rates to achieve high η, as discussed in section 2.5.5. Based on this idea, the PERT solar cell technology evolved, leading to efficiencies of up to 22.5% for large area solar cells, as reported in [42]. It is predicted by the ITRPV that the global market share of PERT solar cells increases significantly in the next decade [5]. Following this trend, n-type Si substrates are used in this work as base material for the development of cost-efficient n-PERT Si solar cells.

When using n-type Si substrates as base material, the major issues are the formation of p-type emitters (usually boron-doped) with high electrical quality, appropriate surface passivation and contact formation. Techniques for the formation of boron emitters and the surface passivation are addressed in the following sections. Contact formation on boron emitters by screen-printing of Ag/Al pastes and its influence on the IV-characteristics of bifacial n-PERT Si solar cells are discussed in chapter 3.

### 2.5 Boron emitter and phosphorus front surface field formation

As mentioned in the previous section, n-type Si is a promising candidate for producing high-efficient Si solar cells because of its beneficial electrical properties compared to the usually used p-type Si. Indeed, by using n-type Si as base material for the fabrication of industrial Si solar cells, one of the key processes is the formation of p⁺-doped emitters. First, an overview of commonly used techniques for the formation of p⁺-doped emitters, usually boron-doped, is presented in the following subsection. Thereafter, an alternative approach for forming boron emitters and simultaneous formation of phosphorus FSF/BSF via sequential co-diffusion from doped borosilicate glasses (BSG), manufactured by means of PECVD, is reviewed. Moreover, the characterization methods used in this work for the quantification of the electrical properties of co-diffused boron emitters and phosphorus FSF/BSF are outlined. At last, the formation of a selective FSF (s-FSF) and its necessity for the minimization of electrical losses attributed to the front surface of the solar cells are briefly discussed.
2.5.1 An overview: Techniques for boron emitter formation

A reliable and well-known approach for the formation of p⁺-doped boron emitters is thermal diffusion of boron atoms from a BSG into the Si substrate. A method for performing boron diffusions for microelectronics was published already in 1978, utilizing a boron tribromide (BBr₃) liquid source [43]. Wang et al. adapted BBr₃-based diffusions to the fabrication of Si solar cells in 1990, demonstrating a solar cell conversion efficiency of 24% [8]. Since then, a lot of research has been conducted on boron diffusions based on BBr₃ [44, 45] or alternatively boron trichloride (BCl₃) [46, 47] liquid sources, however BBr₃ is used more often [5]. The liquid BBr₃ is transported by a carrier gas (usually N₂ or Ar) into a heated quartz tube diffusion furnace and in addition, O₂ is used as reaction gas. BBr₃ and O₂ react to form boron trioxide (B₂O₃), which is deposited all over the surface of the Si wafers. The B₂O₃ reacts with Si to form SiO₂, while elementary boron atoms are released. The mixture of B₂O₃ and SiO₂ is also referred to as BSG. After deposition of the BSG, a drive-in step is conducted, in which the elementary boron atoms diffuse by solid state diffusion into the Si substrate and form the boron emitter. Boron diffusion is usually conducted at temperatures in the range of 900 – 1000°C [44]. Such high temperatures are required because the diffusivity of boron in Si is very low [48]. If the boron concentration in the BSG exceeds a certain level, a boron rich layer (BRL) can form at the wafer surface [49, 50]. The formation of a BRL results in an increased surface recombination activity, which compromises mainly V noct of the solar cell, and either has to be avoided or the BRL has to be removed [50]. The formation of a BRL cannot be attributed to the use of BBr₃, but to boron diffusion in general, if the boron concentration in the BSG exceeds a certain level. This is discussed in more detail in section 2.5.2.

In this work, an alternative approach by means of thermal diffusion of boron atoms from boron containing layers, manufactured by means of chemical vapor deposition (CVD), is used. The CVD technology for the deposition of doped silicon oxide layers was already used in the 1970s for application in microelectronics [51]. Cabal et al. [35] adapted boron-doped silicon oxide (SiOₓ:B) layers, also denoted as BSG, by means of plasma-enhanced CVD (PECVD) to the fabrication of Si solar cells in 2009. Since then, considerable research has been carried out on this topic by several research groups [53-66]. It was shown that SiOₓ:B layers, manufactured by PECVD, can form an appropriate boron diffusion source, resulting in boron emitters with high electrical quality in terms of a low jₑ [62-64]. With this approach, single side doping without the necessity of masking steps is possible, since the SiOₓ:B can be deposited exclusively on one side of the Si substrates. Moreover, only one single high temperature diffusion is required for the formation of boron emitter and phosphorus BSF/FSF, also known as co-diffusion. The co-diffusion and its benefits are discussed in the following subsection.

Note that atmospheric pressure CVD (APCVD) is also a promising technique for the deposition of SiOₓ:B doping layers for application in sequential co-diffusion [58, 59]. Since the deposition is conducted under atmospheric pressure, it provides the potential for further cost reduction. In
addition, boron doping can also be applied by ion implantation [67-69], inkjet printing [70, 71], screen-printing [72, 73] spray-on method [74], spin-on technique [75, 76] and sputtering [37].

2.5.2 Sequential co-diffusion

Figure 2.8a illustrates a frequently used gas-phase based diffusion process flow (e.g. BBr$_3$) [45] for the formation of boron emitter and phosphorus FSF, in comparison to the alternative process flow used in this work for the simultaneous formation of boron emitter and phosphorus FSF using one single co-diffusion step (Figure 2.8b). In the following, these two process flows are compared and the potential of co-diffusion for substantial process simplification is discussed.

First, the n-type Si base material is chemically cleaned after polishing or texturing of the wafer surface in both cases. After pre-treatment of the wafer surface, a capping layer (e.g. SiN$_x$:H by means of PECVD) has to be deposited in case of a typical gas-phase based diffusion process like BBr$_3$, since a gas-phase based diffusion cannot be applied only on one wafer side. This can be explained by the almost uniform deposition of the BSG from the gas-phase over the whole wafer surface, because of the symmetrical arrangement of the wafers in the dopant containing atmosphere in the diffusion furnace. By using a capping layer, the boron emitter solely forms on the uncapped side of the wafer, as illustrated in Figure 2.8a. After this first high-temperature diffusion step, the BSG and possibly the BRL, as well as the capping layer, must be removed by wet chemical etching in an aqueous solution of hydrofluoric acid (HF). Subsequently to chemical etching and wafer cleaning, the boron emitter must be protected by a capping layer (e.g. SiN$_x$:H via PECVD) from overcompensation by phosphorus atoms in the following second high-temperature diffusion step, when forming the FSF during POCl$_3$-diffusion. In a final step, the phosphor silicate glass (PSG) and capping layer must be removed by another wet chemical etching process in HF (5 - 10%).

One drawback of this typical process sequence using gas-phase based diffusions for formation of boron emitter and phosphorus BSF/FSF is the requirement of a total of seven process steps, including two high temperature and masking steps [45]. High temperature treatments are time- and cost-consuming and may also have a detrimental influence, especially on Czochralski-grown (Cz) Si substrates with high oxygen content, so they should be avoided or minimized [9]. Another potential drawback is the depletion of the surface doping concentration $N_S$, which originates from a thermal oxidation step at the end of a typical BBr$_3$-based diffusion [44]. This oxidation step is required for the removal of the BSG and the potentially formed BRL, after typical BBr$_3$ diffusion. The depletion of $N_S$ may result in disadvantages in passivation quality due to increased surface recombination, contact formation and long-term performance stability as shown in [78]. Moreover, it is not possible to decouple the deposition of the BSG and the boron drive-in, since they are conducted successively in one high temperature step.
A promising way to overcome these issues is the use of a sequential *co-diffusion* process as depicted in Figure 2.8b. The BSG is deposited after pre-treatment of the wafer surface by means of PECVD on one side of the wafer. A capping layer (e.g. 120 nm thick SiN$_x$:H via PECVD) is
deposited on top of the BSG to protect the BSG from overcompensation during subsequent POCl₃-diffusion. The capping layer can be deposited in the same PECVD tool directly after deposition of the BSG. The sequential \textit{co-diffusion} is conducted in a two-step process according to Figure 2.9. In the first step, also known as boron \textit{drive-in}, the boron atoms are driven at high temperatures in the range of 900 – 1000°C into the Si substrate. These high temperatures are required because of the low diffusivity of boron in Si. The POCl₃-diffusion is not yet performed, since the diffusivity of phosphorus is much higher than the diffusivity of boron [48]. In the second step, after boron \textit{drive-in} for a duration of 10 – 100 min, depending on the requested emitter profile, a POCl₃-diffusion is performed after cooling down to temperatures commonly used for POCl₃-diffusions (800 - 860°C). The capping layer prevents phosphorus atoms from diffusing through the BSG into the Si substrate and thereby from compensation of the boron doping. In the last step, the PSG, as well as the BSG and capping layer, are simultaneously removed in an aqueous solution of HF (10%). This high concentration of HF is required, since it is difficult to remove SiNₓ:H layers after high-temperature treatments.

The alternative \textit{co-diffusion} process results in a total of four process steps required for formation of boron emitter and phosphorus FSF, compared to a total of seven process steps required when using a common gas-phase based diffusion process. Thus, wafer handling, thermal treatment, process time and process cost can be essentially reduced when using the sequential \textit{co-diffusion}, potentially reducing the LCoE.

Furthermore, \textit{co-diffused} boron emitters do not suffer from depletion of \(N_S\) (as occurring for boron emitters formed by typical BBr₃-based diffusion), enabling high \(N_S\) as discussed in the following. This can be explained by the fact that no thermal oxidation step is required during \textit{co-diffusion} for the subsequent removal of the BSG and potential BRL after \textit{co-diffusion}. The high \(N_S\) may be beneficial for the passivation quality and contact formation in terms of a low contact resistance of screen-printed metal contacts, as discussed in section 3.4.1.

\textbf{Figure 2.9:} Temperature profile of sequential \textit{co-diffusion} for boron emitter and phosphorus BSF/FSF formation in one single high-temperature step. The boron \textit{drive-in} is conducted at high temperatures in the range of 900 - 1000°C, whereas the FSF/BSF formation is performed in a POCl₃-diffusion after cool-down to lower temperatures.
Alternatively, boron- and phosphorus-containing doping layers can be deposited on opposite sides of the wafers prior to co-diffusion. By this means, no gas-phase based diffusion is required at all, as demonstrated in [56]. However, this method results in special requirements for the doping layers and may additionally cause restrictions on the requested doping profiles, since the diffusivity of phosphorus is higher than the diffusivity of boron in Si [48]. Moreover, the doping layers can be structured, enabling selective doping for advanced solar cell concepts like IBC solar cells [55]. Furthermore, the properties of the boron emitters can be adjusted in a more flexible way, because the deposition and diffusion parameters are decoupled [66].

To benefit from the above-mentioned potential of sequential co-diffusion, the co-diffused boron emitters must have a high electrical quality in terms of a low emitter saturation current density $j_0e$ and appropriate contact formation in terms of a low contact resistance, which should be comparable to the ones of boron emitters from typical BBr$_3$-based diffusion. These issues are addressed in section 2.5.4 (electrical quality) and section 3.4.1 (contact formation).

### 2.5.3 Plasma-enhanced chemical vapor deposition

The PECVD technology is a special form of CVD, where the chemical deposition is assisted by a plasma instead of pure thermal dissociation of the reactants. High energetic electrons force the dissociation mechanisms of the reactants and by this means, radicals and ions are generated, which induce the deposition of the layers on the substrate [79]. The temperature of the reaction gases can be kept significantly lower compared to thermal CVD. The plasma source can be located directly at the Si substrate (direct plasma PECVD) or spatially separated (remote plasma PECVD). In the direct plasma method, a strong electrical field is applied between the substrate and a counter electrode, which ignites the plasma. Indeed, the high energetic radicals and ions can damage the Si substrate, degrading the performance of the Si solar cells [80]. Using the remote plasma method, the plasma is located in a separate chamber, therefore plasma-induced damage can be minimized because there is no direct contact between substrate and plasma. A disadvantage of the remote plasma method is the possible loss of radicals on the way from the remote plasma source to the substrate. A promising alternative technology is an inductively coupled plasma (ICP) PECVD, whereby the plasma is induced by an alternating electromagnetic field, obviating the need for electrodes [66]. In this work, direct plasma as well as remote plasma PECVD tools were used as discussed in the following.

The main application of PECVD in this work is the deposition of SiO$_x$:B layers used as doping source in the co-diffusion step for boron emitter formation. As outlined in the previous section, boron emitter formation is commonly realized by thermal diffusion of boron atoms from a solid-state doping source, such as BSG, into the Si substrate at high temperatures commonly in the range of 900 – 1000°C. The BSG is usually formed in a gas-phase based diffusion using BBr$_3$ or BCl$_3$. Alternatively, a SiO$_x$:B (also denoted as BSG) can be deposited using the PECVD technology, which enables single side doping and co-diffusion of boron and phosphorus in one single high temperature step. For the deposition of doped SiO$_x$ layers by means of PECVD, a
Si containing gas, an oxidizing gas as well as a doping gas are required. For the deposition of a BSG, most commonly (also in this work), silane (SiH₄), carbon dioxide (CO₂) or nitrous oxide (N₂O) and diborane (B₂H₆:H) diluted in hydrogen are used. Cabal et al. [35] used tetraethylorthosilicate (Si(OC₂H₅)₄) and triethylboron ((C₂H₅)₃B) as alternative precursor gases. In [45], trimethylborate (B(OCH₃)₃) was used as doping gas instead of diborane.

The deposition of the BSG can be conducted at low temperatures with a deposition time of some minutes [35]. Thus, no essential diffusion of the doping atoms into the Si substrate occurs during deposition. Therefore, the parameters of the deposition are completely decoupled from the parameters of the diffusion (respectively drive-in step). This enables flexible adjustment of the boron emitter properties and allows a wider variation of the doping profiles, compared to common gas-phase based boron diffusions [66]. The surface doping density \( N_S \) of the boron emitter and shape of the doping profile near the wafer surface depend mainly on the properties of the doping source, such as the BSG thickness and boron concentration [35]. In turn, the BSG thickness and the boron concentration depend strongly on the reaction gases used, gas flows (especially of the doping gas), gas pressure, deposition temperature, deposition duration and applied PECVD method (direct or remote plasma). The emitter depth \( d_e \) depends mainly on diffusion parameters, such as temperature and duration of the diffusion. Engelhardt et al. [66] reported that the boron concentration in the BSG also has a substantial influence on \( d_e \). The study was conducted using the same low-pressure single-chamber ICP-PECVD lab tool used in this work. Four large area (243.36 cm²) wafers can be deposited simultaneously. The plasma is ignited by an inductively-coupled alternating electromagnetic field with a frequency of 13.56 MHz. SiH₄, CO₂ and B₂H₆:H were used as reaction gases for the deposition of the BSG. Engelhardt et al. demonstrated, that the boron concentration in the BSG increases continuously for increasing diborane gas flow, while keeping all other deposition parameters constant. The boron concentration in the BSG was measured to be in the range of 0.5 – 1.8 \times 10^{22} \text{ cm}^{-3} for a diborane flow in a range of 400 – 1000 sccm. They observed, that \( d_e \) increases at first from 0.35 μm to 0.55 μm for an increasing diborane flow from 400 sccm to 700 sccm, while keeping the diffusion duration and temperature constant (100 min at 930°C). This trend is reversed for higher diborane flow, assisting the formation of a BRL, due to the high boron concentration in the BSG. They assumed that the BRL prevents boron atoms from diffusing into the Si substrate, thereby reducing \( d_e \) from 0.55 μm to 0.45 μm for an increasing diborane flow from 700 sccm to 1000 sccm. Furthermore, they observed that the boron profile is depleted for low diborane flows, since the boron concentration in the BSG is too low, in this case, to act like an infinite doping source. Thus, boron atoms diffuse back into the BSG, which has a higher solubility limit for boron compared to Si, during cooldown.

Moreover, Engelhardt et al. demonstrated in [46] that ICP-PECVD BSG layers can also be applied multifunctionally, simultaneously acting as doping layer, passivation layer and ARC. They showed that the refractive index of the BSG is around 1.5 at 633 nm, slightly increasing with increasing boron concentration, and that the absorption coefficient of the BSG is below
It turned out that a BSG (thickness of 76 nm) stacked with a SiN$_{x}$:H layer (thickness of 57 nm) can even improve the optical properties of the solar cells, compared to a single SiN$_{x}$:H ARC with a thickness of 75 nm. In addition, the BSG can also be used as passivation layer, which exhibits electrical properties comparable to those of an Al$_2$O$_3$/SiN$_{x}$:H passivation layer stack (see section 2.6.2), when using optimized deposition and diffusion parameters. Despite these promising results, the BSG was exclusively used as doping layer in this work.

It has to be mentioned that for some of the experiments presented in this work, the PECVD tool Oxford Plasmalab System 100 from the manufacturer Oxford Instruments was used. This tool uses a direct plasma source with a plasma frequency of 13.56 MHz. Using this tool, N$_2$O was used as the oxidizing gas instead of CO$_2$. Only one large-area wafer can be deposited in the process chamber and plasma damage cannot be avoided. Therefore, the ICP-PECVD tool was preferably used for the deposition of the BSG.

The second application of PECVD is the deposition of SiN$_{x}$:H layers which can also be used multifunctionally, providing protection against chemical etching, diffusion barriers, passivation layers and/or ARC [81]. Usually, SiN$_{x}$:H layers are deposited by PECVD technology, using SiH$_4$ and ammonia (NH$_3$) as reaction gases. One of the multifunctional applications is the use as ARC, e.g. as an Al$_2$O$_3$/SiN$_{x}$:H layer stack commonly used for passivation of boron emitters as discussed in section 2.6.2. Moreover, the SiN$_{x}$:H layers show excellent properties as capping layers against several chemical treatments, such as texturing and polishing, e.g. in an aqueous solution of potassium hydroxide (KOH). In this work, the SiN$_{x}$:H layers are additionally used as capping layers against indiffusion of phosphorus atoms during sequential co-diffusion in a POCl$_3$ quartz tube furnace. To protect the BSG from overcompensation by phosphorus atoms, a SiN$_{x}$:H layer with a thickness of 120 nm is deposited on top of the BSG before co-diffusion. The use of SiN$_{x}$:H layers does not result in a depletion of $N_S$, in contrary to the use of SiO$_x$ capping layers, even when they are directly deposited on top of the boron emitters [82]. It turned out that an ICP-PECVD SiN$_{x}$:H layer with a thickness of 20 nm already suffices as a capping layer against typical POCl$_3$ diffusion [81]. The capping properties of the SiN$_{x}$:H layers depend essentially on the density of the layers and therefore on the PECVD tool used. SiN$_{x}$:H layers deposited by the remote PECVD tool SiNA XS from the manufacturer Roth & Rau did not show a sufficient barrier effect in the experiments presented in this work.

### 2.5.4 Characterization of emitter properties

It was mentioned in the previous subsections that sequential co-diffusion from doping layers manufactured by means of PECVD is a promising method for simultaneous formation of boron emitter and phosphorus FSF in one single high temperature diffusion. This method enables considerable process simplifications compared to typical gas-phase based diffusion processes. However, the use of sequential co-diffusion is only beneficial if a high electrical quality of the co-diffused boron emitters, in terms of a low emitter saturation current density $j_{0e}$, and good contact formation can be achieved.

10$^{-4}$.
A highly-doped region (emitter or FSF/BSF) can be characterized by parameters such as the surface doping concentration $N_S$, depth of the highly-doped region $d$, sheet resistance $R_{sh}$ and saturation current density $j_0$, which accounts for all recombination mechanisms in the highly-doped region and at the wafer surface, as outlined in section 2.3.2.

Electrochemical capacitance-voltage (ECV) measurements [83] are performed to determine the electrically active doping concentration of the highly-doped region as a function of depth $N_D(z)$, also referred to as doping profile. $R_{sh}$ can be quantified by 4-point probe (4PP) measurements or calculated from the doping profile by integration over the depth of the whole doping profile

$$\frac{1}{R_{sh}} = q \int \mu_p/n(z) \cdot N_D(z) \, dz,$$

(2.30)

with the mobilities of holes $\mu_p$ and electrons $\mu_n$, which depend on the doping concentration and are therefore also a function of depth. Prerequisite for ECV measurements is a Schottky-contact between metal and Si substrate, which generates an electrical field. The Schottky-contact acts like a parallel-plate capacitor. The width of the SCR can be influenced by applying a voltage $V$. To measure the capacity $C$, the applied direct voltage is overlaid with a weak alternating voltage, enabling the doping concentration to be calculated from the measured $C$-$V$ curve.

Figure 2.10 shows an example of the variations in doping profiles of co-diffused boron emitters for an optimized deposition process and different diffusion parameters using the PECVD tool Oxford Plasmalab System 100 (published in [60]). The difference in the drive-in parameters for the boron emitter with $R_{sh} = 41$ $\Omega$/sq (ECV), compared to the boron emitter with $R_{sh} = 54$ $\Omega$/sq (ECV) is only the diffusion duration, while using the same diffusion temperature of 930°C. For the boron emitter with $R_{sh} = 92$ $\Omega$/sq (ECV), the boron drive-in step was conducted at a lower diffusion temperature, resulting in lower $N_S$, as well as lower $d_e$. This can be explained by the lower solubility limit of boron in Si for lower temperatures on the one hand and by the lower diffusivity of boron in Si for lower temperatures on the other hand [48]. Because the doping concentration was high enough to act as an infinite doping source in all cases, no depletion of $N_S$ is observed. Therefore, $N_S$ of up to $10^{20}$ cm$^3$ can be achieved for a diffusion temperature of 930°C. It is visible that the presented boron emitters also do not suffer from formation of a BRL, due to optimized deposition and diffusion parameters. As reference, the doping profile of a boron emitter fabricated by a common BBr$_3$-based gas-phase diffusion with $R_{sh} = 54$ $\Omega$/sq (ECV), suffering from substantial surface depletion due to thermal oxidation, is also shown. No noticeable difference in the shape of the doping profile (except for the depletion of $N_S$) are observed between the BBr$_3$-based reference boron emitter and the co-diffused boron emitter (compare blue and gray line in Figure 2.10).

However, the shape of the doping profile is not the only requirement for optimal boron emitter formation by means of co-diffusion. The co-diffused boron emitters must also exhibit a high electrical quality in terms of low $j_{0e}$ and high implied $V_{oc}$ ($iV_{oc}$) values to enable high efficiency of the n-PERT solar cells.
Figure 2.10: ECV doping profiles of co-diffused boron emitters from BSG by means of PECVD with sheet resistivities in the range of 50 – 90 Ω/sq. The doping profile of a boron emitter fabricated by means of typical BBr₃-based diffusion is shown as reference. Published in [60].

Kane and Swanson developed a fast and reliable method for quantifying \( j_{0e} \) by a contactless photoconductance decay method [25]. The mean lifetime \( \tau \) of the minority charge carriers has to be known as a function of the injection level \( \Delta p \). Depending on its value, the \( \tau(\Delta p) \) curve can be determined inductively by the transient photoconductance decay method which is induced by a short flash or by the quasi-steady-state (QSS) photoconductance method using a long flash of a flashlight [84, 85].

The transient photoconductance decay method is only valid for \( \tau > 200 \) μs. In this work, the Sinton WCT-120 lifetime tester was used for the lifetime measurements [85]. If \( \tau \leq 200 \) μs, the QSS photoconductance method has to be used instead [86, 87].

The inverse of the minority carrier lifetime \( \tau \) can be defined according to equation (2.27) as

\[
\frac{1}{\tau_{eff}} = \frac{1}{\tau_{Auger}} + \frac{1}{\tau_{SRH}} + \frac{2j_0(N_D + \Delta p)}{qWn_i^2} \tag{2.31}
\]

Note that equation (2.31) is only valid for symmetrical samples with equal saturation current density \( j_0 \) of front and rear side. Therefore, symmetrical test structures were prepared in this work according to Figure 2.11. The bulk of the wafers should be of high electrical quality, so that the SRH recombination in the bulk can be neglected. Thus, FZ-Si wafers with a high base resistivity of 200 Ωcm are usually used for \( j_{0e} \) measurements. The SRH recombination is also very low for the high-ohmic n-type Cz Si wafers used in this work. Moreover, the resistivity of the base material has to be very high in order to ensure that the base is in high injection (\( \Delta p > 10N_D \)) during measurement. This leads to injection-independent SRH recombination.
The Auger recombination must be calculated by an adequate parameterization, such as those presented in [88-91].

The method of Kane and Swanson for the determination of $j_{0e}$ is based on the calculation of the slope of the Auger corrected inversed lifetime curve (equation (2.31)), while neglecting the SRH recombination. The calculation of $j_{0e}$ from the inversed lifetime curve was performed at an injection level $\Delta p = 5 \times 10^{15} \text{cm}^{-3}$, assuming $n_i = 8.6 \times 10^9 \text{cm}^{-3}$ at room temperature [92]. The injection level corresponds to the operation condition under an illumination of 1 sun for the high-ohmic n-type Cz wafers.

From the $j_{0e}$ data the $iV_{oc}$ values can be calculated according to equation (2.3)

$$iV_{oc} = \frac{kT}{q} \cdot \ln\left(\frac{j_{ph}}{j_{0e}} + 1\right),$$

(2.32)

or by the excess concentration of charge carriers [93]

$$iV_{oc} = \frac{kT}{q} \cdot \ln\left(\frac{\Delta p(N_D + \Delta p)}{n_i^2} + 1\right).$$

(2.33)

Table 1 shows the sheet resistance $R_{sh}$, which was measured by a 4PP and ECV setup, emitter saturation current density $j_{0e}$ and $iV_{oc}$ values of symmetrical test structures for the co-diffused boron emitters shown in Figure 2.10 and common BBBr3-based gas-phase diffusions (published in [60]). The boron emitters were passivated with an Al$_2$O$_3$/SiN$_x$:H passivation layer stack to ensure ideal surface passivation (see section 2.6.2). The 4PP measurements revealed $R_{sh}$ values in the range of 51 – 94 $\Omega$/sq, while ECV measurements differ slightly, revealing $R_{sh}$ values in the range of 41 – 92 $\Omega$/sq. $j_{0e} < 40$ fA/cm$^2$ can be achieved for the co-diffused boron emitters, enabling $iV_{oc} > 690$ mV. This is comparable to the values achieved for optimized BBBr3-based diffusions. This confirms the high electrical quality of the co-diffused boron emitters.

In a next step, the industrial feasibility of sequential co-diffusion must be demonstrated on large-area (up to 243.36 cm$^2$) industrial n-type Si solar cells with $\eta$ comparable to the ones of solar cells with a gas-phase based process sequence. Therefore, the IV-characteristics of co-diffused bifacial/mono-facial n-PERT solar cells fabricated in this work are presented in chapter 3/4.
Table 1: Sheet resistance $R_{sh}$, measured by 4PP and ECV, emitter saturation current densities $j_0e$ and implied $V_{oc}$ values for different boron emitters by means of co-diffusion from PECVD BSG and common BBr$_3$-based gas-phase diffusions with Al$_2$O$_3$/SiN$_x$:H passivation. Published in [60].

<table>
<thead>
<tr>
<th>boron emitter type</th>
<th>$R_{sh}$ (4PP) [Ω/sq]</th>
<th>$R_{sh}$ (ECV) [Ω/sq]</th>
<th>$j_0e$ [fA/cm$^2$]</th>
<th>$iV_{oc}$ [mV]</th>
</tr>
</thead>
<tbody>
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<td>PECVD</td>
<td>51</td>
<td>41</td>
<td>61</td>
<td>680</td>
</tr>
<tr>
<td>PECVD</td>
<td>62</td>
<td>54</td>
<td>44</td>
<td>688</td>
</tr>
<tr>
<td>PECVD</td>
<td>94</td>
<td>92</td>
<td>37</td>
<td>692</td>
</tr>
<tr>
<td>BBr$_3$</td>
<td>50</td>
<td>-</td>
<td>45</td>
<td>687</td>
</tr>
<tr>
<td>BBr$_3$</td>
<td>100</td>
<td>-</td>
<td>32</td>
<td>696</td>
</tr>
</tbody>
</table>

2.5.5 Front surface field and selective etch-back

Although POCl$_3$-based diffusion is most frequently used for the formation of phosphorus-doped emitters of p-type Si solar cells, it can also be used for the formation of highly doped FSF/BSF of n-type Si solar cells. The working principle, formation and passivation of a s-FSF fabricated by means of selective chemical etch-back (EB) are presented in the following.

The n$^+$-doped FSF has to fulfill several requirements simultaneously such as efficient shielding of the front surface and metal contacts by rejecting minority charge carriers, which results in low recombination rates (low $j_0$). This is essential for high-efficient RJ Si solar cell concepts since the majority of charge carriers is photo-generated close to the front surface. Furthermore, appropriate contact formation by screen-printing in terms of a low contact resistance between metal fingers and FSF $R_{c,f}$ and low lateral resistivity $R_{FSF}$ must be fulfilled by the FSF. $R_{c,f}$ and $R_{FSF}$ have to be kept as low as possible to minimize fill factor $FF$ losses induced by series resistance $R_S$ (see section 2.3.3).

$R_{FSF}$ of the n$^+$-doped FSF for n-type Si solar cells is not as critical as for p-type Si solar cells since it has the same polarity as the n-type base. Therefore, the n-type base contributes to lateral charge carrier transport itself, resulting in an effective sheet resistance $R_{sh,eff}$ of base and FSF, which can be calculated according to equation (2.30) by integration over the FSF and base to the pn-junction. Due to the high base resistivity ($\rho_b = 6 – 10 \text{ Ωcm}$) of the n-type Si substrates used in this work, which corresponds to a low doping concentration ($N_D < 10^{15} \text{ cm}^{-3}$), the photo-generated charge carriers must also be considered for lateral carrier transport. This results in an injection level dependent effective base resistivity $\rho_{b,eff}(\Delta p)$, as reported by Book et al. in [94] using PC1D simulation [95, 96]. The simulation revealed $\rho_{b,eff}$ of 2.8 - 3.4 Ωcm at MPP for the n-type Si base substrates used in this work. Based on this data, $R_{sh,eff}$ of the FSF and base, as a function of $\rho_{b,eff}$, was calculated according to equation (2.30) for different sheet resistivities of the FSF $R_{sh,FSF}$ (see Figure 2.12).
It is evident that $R_{sh,eff}$ is essentially lower than $R_{sh,FSF}$ since the base is already in high injection ($\Delta p > N_D$) at MPP and the generated charge carriers contribute considerably to lateral charge carrier transport. Hence, the effective lateral resistance of a $n^+$-doped FSF using $n$-type base substrates is lower than the lateral resistance of $n^+$-doped emitters using $p$-type substrates. This must be considered for the optimization of the front side metallization $H$-grid pattern at MPP (see section 3.3.5).

As reported in [97], high surface doping concentrations and deep doping profiles of the FSF in the illuminated cell area result in noticeable losses of $j_{sc}$ and $V_{oc}$ because of the limiting Auger recombination, which is directly correlated to the doping concentration (see section 2.3.2). This behavior is almost independent of the surface recombination velocity for high surface doping concentrations and deep emitter profiles. A low surface doping concentration, on the other hand, reduces the field-effect of the FSF which compromises mainly $V_{oc}$ if the surface is not well passivated [97]. In addition, the as-diffused phosphorus doping profile suffers from a dead layer close to the surface with high recombination rate, inducing a high saturation current density and substantial loss in the short-wave range of the incident light. The dead layer can be removed by wet-chemical etching of several nm of the doping profile, known as emitter EB process [98]. The EB is based on an aqueous solution of HF and nitric acid ($\text{HNO}_3$).

Since a highly-doped region is beneficial beneath the metal contacts in terms of a good electrical contact with low $R_{cf}$ (see section 3.3.3) and excellent shielding of the metal contacts which lowers recombination at the metal contacts, the formation of a s-FSF has been developed at the University of Konstanz using a selective EB process [98-102]. The highly-doped region can be preserved by selective masking with an etch-resistant mask, whereas the dead layer is removed in the non-masked regions. Thus, the electrical losses on the front surface can be minimized.

---

**Figure 2.12:** Effective sheet resistance $R_{sh,eff}$ of FSF and base as a function of the effective base resistivity $\rho_{b,eff}$ for different sheet resistivities of the FSF at MPP. The lines are guides to the eye.
because the doping profiles in the illuminated and contacted areas can be adjusted almost independently of each other. Selective masking can be realized by opening full-area etch-resistant masks with laser or etching paste [57], screen-printing of etch-resistant pastes or inkjetting of etch-resistant inks [102].

In this work, a POCl₃ diffusion is conducted during sequential co-diffusion (see section 2.5.2), endeavoring to achieve a sheet resistance $R_{sh} \approx 30 - 40 \, \Omega/\text{sq}$. After co-diffusion and selective masking by means of inkjet printing of an etch-resist ink using the inkjet tool DoD-300 from the manufacturer Schmid, the wafers are etched-back to $R_{sh} \approx 70 - 100 \, \Omega/\text{sq}$. The wafer surface gets porositized by the EB etching solution showing a characteristic color in dependence of the thickness of the porous Si. First the FSF is etched-back on test wafers (p-type Si) with the same phosphorus doping profile as the solar cells until the requested $R_{sh}$ is reached. The color of the porous Si is used as reference. Subsequently, the solar cells are etched-back until the porous Si shows the reference color, confirming the required etching depth. Note that the required EB time depends essentially on the portion of nitrogen monoxides (NO) in the EB solution and can be adjusted by the addition of sodium nitrite (NaNO₂) of up to 1.25 g/l [97].

The BSG and capping layer can be used for protection of the boron emitter against the EB solution. However, the potentially formed BRL during co-diffusion must be removed to achieve a low emitter saturation current density $j_0e$, which is mandatory when targeting n-PERT solar cells with high $\eta$. The BRL can be removed during selective EB of the FSF [50]. Furthermore, it turned out in this work that it is beneficial to completely remove the BSG and capping layer before EB to remove the boron doping at the edges of the wafers and the potential wraparound of the BSG on the front surface. The wraparound of the BSG leads to boron doping at least at the edges of the wafer, which results in bad edge isolation, causing a low parallel resistance $R_P$, but may also cause weak boron doping on the front surface, which may result in short-circuits induced by the front metallization in this area. Its detrimental impact on the IV-characteristics of n-PERT solar cells can be lowered when conducting the EB after removal of the BSG and capping layer. However, the etching rate is usually higher for p-doped Si like boron emitters compared to n-doped Si like the phosphorus FSF, since the chemical EB relies on positive charge carriers. It was observed in this work that the EB results in a substantial increase of $R_{sh}$ of the lowly doped boron emitter without BRL, while $R_{sh}$ only slightly increases for the highly doped boron emitter with BRL, since the etch rate of the BRL is very low [50].

After EB, the porous Si and etch-resistant mask are removed simultaneously in an aqueous solution of KOH (< 0.4%) and diethylenglycolmonobutylether (BDG, < 7%), using an ultrasonic bath. The KOH removes the porous Si and enhances the dissolving of the etch-resistant mask, whereas the BDG also induces the dissolving of the etch-resistant mask but additionally regulates the etching-rate of the KOH. More information on the etching process and chemical etching reactions can be found in [97, 103, 104]. It was observed in this work that KOH/BDG solution strongly affects $R_{sh}$ of the FSF and boron emitter for long etching times. Therefore, the duration of the KOH/BDG treatment has to be minimized to avoid unnecessary etching of the
doping profiles, resulting in an increase or generation of lateral inhomogeneities of the doping. It turned out in the experiments, an etching time of around 120 s was required for sufficient removal of the etch-resist mask.

If the boron emitter is protected by the BSG and SiNₓ:H capping layer during EB, they must be removed afterwards in HF (10%). This etching step results only in a negligible increase of $R_{sh}$ of the phosphorus FSF. Furthermore, the thermal oxidation for 5 min at 900°C for the formation of the SiO₂ passivation layer on the s-FSF influences the doping profiles considerably. It was observed that the as-diffused $R_{sh}$ of the boron emitter must be in the range of 65 Ω/sq if a target $R_{sh}$ of around 80 Ω/sq is requested (in the case of a protected boron emitter during EB). A boron emitter with even lower $R_{sh}$ must be diffused for the preferred case when the boron emitter is not protected by the capping layer during EB. The phosphorus FSF has to be etched-back from its initial value to $R_{sh}$ of around 75 Ω/sq if a post-process $R_{sh}$ of 100 Ω/sq is requested.

Therefore, the deposition and diffusion parameters must be adjusted, considering the increase of $R_{sh}$ of the boron emitter during formation of the s-FSF via EB and thermal oxidation.

### 2.6 Surface passivation

At the surface of crystalline Si substrates, the crystal structure is defective, inducing a high density of dangling bonds, which act as recombination centers for charge carriers (compare section 2.3.2). Foreign atoms located at the Si surface induce additional recombination centers. This leads to significant power losses of the solar cells attributed to an enhanced current density of the first diode $j_{01}$ (compare section 2.3.4). Therefore, the surface has to be passivated in order to minimize recombination losses and improve the conversion efficiency $\eta$ of the solar cells.

Two principles for surface passivation can be differentiated. The first principle is based on the reduction of the density of defect states at the wafer surface, known as chemical passivation. This can be achieved by chemical saturation of the dangling bonds due to the passivation layer itself or by atomic hydrogen which can diffuse from the hydrogen-rich passivation layer to the wafer surface during contact firing [105]. The second principle is based on electrostatic shielding of the wafer surface by fixed charges located in the passivation layer or doping gradients, known as field effect passivation. The polarity of the fixed charges has to match the polarity of the minority charge carriers in order to effectively hinder them from reaching the wafer surface. Note that an unfitting polarity of the fixed charges can also lead to good surface passivation, if an inversion layer forms underneath the passivation layer for a high density of fixed charges. However, this inversion layer leads to parasitic currents if the solar cell is locally contacted in this area, also referred to as parasitic shunting [106]. Therefore, different passivation layers are usually used for p- and n-doped Si surfaces.

For n-doped regions like phosphorus emitters, surface passivation is typically realized using SiNₓ:H passivation layers deposited by PECVD, as introduced in section 2.5.3. The SiNₓ:H
layers contain a high density of positive fixed charges, which reject the minority charge carriers in the n-doped region, hindering them from reaching the wafer surface [61]. For the majority of industrial p-type Si solar cells, a full-area Al BSF is formed for field effect passivation, leading to a moderate passivation quality (surface recombination velocity \( s \approx 200 - 600 \text{ cm/s} \)) [108]. This leads to a limitation of the solar cell conversion efficiency to around 20%. With the application of n-type Si as base material for industrial solar cells, alternative passivation layers must be used since SiNx:H layers are not well-suited for the passivation of emitters with p-type doping [63], partly caused by the inappropriate polarity of the fixed charges.

In the following sections, the passivation layers used within this work are introduced. First, a well-established method in laboratories for the passivation of p-doped Si using SiO\(_2\) by thermal oxidation [110-112], is discussed. Thereafter, an alternative approach for the passivation of p-doped emitters by deposition of Al\(_2\)O\(_3\) at low temperatures [113-124] is presented.

### 2.6.1 Thermal oxidation

It is known that thermal oxidation can provide a good chemical passivation by strong reduction of the density of interface defects on p- and n-doped Si surfaces, whereas field effect passivation plays only a minor role due to the low density of positive fixed charges [110-112]. Therefore, thermally grown SiO\(_2\) layers can be applied for p- and n-doped Si surfaces. The thermal dry oxidation is usually conducted at temperatures in the range of 800 – 1000°C in a quartz tube furnace in oxygen ambient [110]. The SiO\(_2\) grows into the Si wafer, leading to a strongly decreased density of defect states at the Si/SiO\(_2\) interface. Because thermal SiO\(_2\) has a refractive index of around 1.5, it cannot be used as ARC (see section 2.3.1). To overcome this issue, a thin SiO\(_2\) layer of 5 – 15 nm stacked with a 60 -70 nm thick hydrogen rich SiNx:H layer was used for surface passivation of the FSF/BSF of the PERT solar cells investigated in this work. The thicker SiNx:H layer dominates the optical properties of the passivation stack, whereas the thin SiO\(_2\) layer facilitates the high quality of surface passivation. In addition, the passivation quality is enhanced by thermal diffusion of hydrogen from the SiNx:H layer to the Si/SiO\(_2\) interface and into the Si wafer during contact firing, which further reduces the density of defect states at the Si/SiO\(_2\) interface and in the Si bulk [105].

Note that silicon oxide can also be deposited by PECVD [125] and grown chemically, e.g. by a nitric acid pretreatment [126] etc.

### 2.6.2 Atomic layer deposition of Al\(_2\)O\(_3\)

Studies on thermal atomic layer deposition (ALD) of Al\(_2\)O\(_3\) revealed surface recombination velocities \( s < 10 \text{ cm/s} \) on p-doped Si substrates. This can be explained by a high density of fixed negative charges and a low density of defects at the Si/Al\(_2\)O\(_3\) interface [114, 115]. Excellent results were also reported with plasma-assisted ALD (PA-ALD), providing excellent surface passivation, especially on p\(^+\)-doped surfaces [116, 119]. It was shown in [119] that the use of
PA-ALD leads to a higher density of fixed negative charges compared to thermal ALD. The high density of fixed negative charges and the low density of interface defects of PA-ALD Al$_2$O$_3$ layers enable the passivation of boron emitters with high electrical quality.

The PA-ALD technique is based on two reactants which are injected alternately in the reaction chamber and each of them reacts with the surface of the substrate in a self-limiting reaction [127]. The reactants are trimethylaluminum (C$_6$H$_{18}$Al$_2$) and O$_2$. In this work, the PA-ALD FlexAl II manufactured by Oxford Instruments was used for the deposition of thin Al$_2$O$_3$ layers. Because the deposition is performed at temperatures $\leq 300^\circ$C, no degradation of $\tau$ in the bulk due to thermal treatment is observed (compared to thermal oxidation outlined in the previous section). Thermal activation for 30 minutes in a N$_2$ atmosphere at 420°C is required to reach an optimum of passivation quality of the Al$_2$O$_3$ passivation layers [116, 124]. The improvement in passivation quality by thermal activation can be attributed to hydrogenation of the interface by diffusion of atomic hydrogen originating from the Al$_2$O$_3$ layer [123]. For optimized deposition parameters, $j_{0e} \approx 10$ fA/cm$^2$ on BBr$_3$ boron emitters with $R_{sh} > 100$ Ω/sq can be achieved [128]. The Al$_2$O$_3$ layers must be thin to avoid delamination during high temperature treatments [124], such as the contact firing which is necessary for solar cell production. Although thermal activation at 420°C for 30 min is optimal, it is reported in [121] that contact firing also results in an appropriate activation.

Since Al$_2$O$_3$ has a refractive index $n < 1.7$, it cannot be used as ARC (see section 2.3.1). Thus, a thin Al$_2$O$_3$ layer of 10 nm (97 cycles at 300°C) stacked with a thick hydrogen rich SiNx:H layer (thickness of 65 nm) was used for surface passivation of the boron emitters of the n-PERT cells in this work. Note that Al$_2$O$_3$ can also be deposited by PECVD [117], APCVD [113, 120], sputtering [118] and pyrolysis [129].

### 2.6.3 Influence of pre-treatment on passivation quality

The influence of wafer storage time on the passivation quality of PA-ALD Al$_2$O$_3$ passivation layers on n-type Si base material and boron emitters after chemical cleaning was studied in this work on symmetrical lifetime samples (without emitter) and symmetrical $iV_{oc}$ samples (with 80 Ω/sq boron emitters fabricated by BBr$_3$-diffusion) using n-type float-zone (FZ)-Si wafers with $\rho_b \approx 10$ Ωcm and $W \approx 250$ µm. Furthermore, the influence of the storage time after ALD before deposition of the SiNx:H layer was studied. No chemical polish is required, when using FZ-Si wafers, because these wafers are already chemically polished by the manufacturer. The wet chemical cleaning process used for removing of metallic and organic contamination from the wafer surface is also known as RCA cleaning. The first step of the RCA cleaning takes place in an aqueous solution of NH$_3$ and hydrogen peroxide (H$_2$O$_2$). This solution oxidizes the wafer surface due to H$_2$O$_2$ and the grown thin layer of chemical silicon oxide, containing mainly the organic contaminations from the wafer surface, is removed in an aqueous solution of HF (2%). The second step involves treating the wafers in an aqueous solution of hydrochloric acid (HCl).
and H$_2$O$_2$. The thin chemical silicon oxide, formed during this cleaning step, containing mainly the metallic contaminations, is again removed in an aqueous solution of HF (2%).

The chemical cleaning was performed simultaneously for all wafers. Subsequently, the wafers were stored in a clean box for different time intervals over a range of 0 – 192 h before ALD. The Al$_2$O$_3$ passivation layers were deposited on both wafer sides via PA-ALD at 170°C for 250 cycles (≈ 25 nm Al$_2$O$_3$), as well as at 300°C for 97 cycles (≈ 10 nm Al$_2$O$_3$). Note that for the symmetrical samples with boron emitter the ALD was solely conducted at 300°C and the Al$_2$O$_3$ layers were capped with a SiN$_x$:H layer with a thickness of 65 nm, corresponding to the process sequence used for the fabrication of bifacial n-PERT Si solar cells, as discussed in section 3.1. To activate the passivation, the wafers were annealed for 30 min at 420°C in a nitrogen atmosphere (deposition at 170°C) or fired in an IR belt furnace at a set peak firing temperature $T_S = 800°C$ (deposition at 300°C). The firing step corresponds to the contact firing step for the fabrication of n-PERT solar cells. The measured minority charge carrier lifetime $\tau$ after thermal annealing is shown in Figure 2.13, whereas Figure 2.14 illustrates the $iV_{oc}$ values for different storage times after RCA cleaning (x-axis) and different storage times after ALD of the Al$_2$O$_3$ layers before the PECVD of the SiN$_x$:H layers (y-axis). No substantial difference is observed for the different time intervals of wafer storage between deposition directly after the chemical cleaning (0 min) and up to 192 h storage time. The measured average lifetime of around 5/10 ms corresponds to a surface recombination velocity $S = 2.5/S = 1.25$ cm/s. This and the high $iV_{oc}$ values of up to 690 mV confirm the excellent passivation quality of the PA-ALD Al$_2$O$_3$ layers.

![Figure 2.13: Minority charge carrier lifetime of polished n-type FZ-Si wafers (10 Ωcm): a) 25 nm thick PA-ALD Al$_2$O$_3$ passivation layers deposited at 170°C and annealed at 420°C for 30 min in a nitrogen ambient and b) 10 nm thick PA-ALD Al$_2$O$_3$ passivation layers deposited at 300°C stacked with 65 nm thick SiN$_x$:H layers with/without firing step in an IR belt furnace at a set peak firing temperature $T_S$ of 800°C. The different time data corresponds to the storage time after chemical cleaning.](image-url)
Note that the same studies were performed for alkaline textured wafers. This data is not shown since residues from the alkaline texture are expected to have remained on the wafer surface, even after RCA cleaning, resulting in a very high $S$. However, the same influence of the storage time on the passivation quality observed on polished wafers is also expected on textured wafers.

In a further experiment, it turned out that KOH-related residues cannot be sufficiently removed by RCA cleaning, but by a wet chemical cleaning process, known as *piranha* cleaning. The *piranha* cleaning is conducted in a solution of sulfuric acid (H$_2$SO$_4$) and H$_2$O$_2$. The thin oxide layer forming during the cleaning process can be removed in an aqueous solution of HF (2%). Therefore, *piranha* cleaning was preferred for wafer cleaning during the fabrication of n-PERT solar cells.

![Figure 2.14: Contour plot of $iV_{oc}$ of symmetrical samples with BBr$_3$-based boron emitters with $R_{sh} = 80$ $\Omega$/sq for different storage times after RCA cleaning (x-axis) and for different storage times after ALD of the Al$_2$O$_3$ passivation layer at 300°C before PECVD of the SiN$_x$:H layer (y-axis). The samples were fired at 800°C.](image)

Based on these findings, it is assumed that neither the storage time after RCA cleaning nor the storage time after ALD before PECVD have a noticeable impact on the passivation quality of PA-ALD Al$_2$O$_3$ passivation layers on base material as well as on boron emitters, if the wafers are stored in a clean box which prevents the wafers from collecting dust.

### 2.7 Summary

The outlined power loss mechanisms and their limiting influence on the IV-characteristics of Si solar cells are the basis for understanding, quantifying and minimizing these loss mechanisms for improving the fabrication processes of n-PERT Si solar cells performed in this work. The
metallization-induced power loss of the solar cells can have a considerable contribution to the total power loss of the solar cells.

The main issues when using n-type Si as base material are the formation of p⁺-doped emitters and sufficient surface passivation. Simultaneous formation of boron emitter and phosphorus FSF by means of sequential co-diffusion from doping layers manufactured by PECVD enables a simplified process sequence for potentially time- and cost-efficient fabrication of industrial n-PERT solar cells, compared to usually used gas-phase based diffusions like BBr₃. Sequential co-diffusion enables a high flexibility in the boron emitter profiles, since the deposition and diffusion parameters are completely decoupled. High surface doping concentrations of up to \( N_S = 10^{20} \text{ cm}^{-3} \) without depletion can be achieved, while maintaining high \( iV_{oc} > 690 \text{ mV} \) and low emitter saturation current densities \( j_{0e} < 40 \text{ fA/cm}^2 \), using an Al₂O₃/SiNx:H passivation layer stack. These results confirm the high electrical quality of co-diffused boron emitters.

The formation of a s-FSF by selective EB is required for minimizing power loss at the front surface. The saturation current density of the front surface in the illuminated areas is reduced to \( j_{0FSF} < 60 \text{ fA/cm}^2 \) using a SiO₂/SiNx:H passivation layer stack, while retaining high doping underneath the metal contacts, which is beneficial for electrical shielding of the metal contacts and optimal contact formation in terms of a low contact resistance. The EB process also affects the boron profile in the absence of a BRL, while the boron emitter profile is only slightly etched in the presence of a BRL due to the low etch rate of the BRL. Therefore, the deposition process and diffusion parameters must be adjusted to achieve boron emitters with the requested \( R_{sh} \).
3 Metallization-Induced Power Loss of Bifacial n-Type Silicon Solar Cells

As mentioned in the previous chapter, n-type Si as base material for the fabrication of Si solar cells has the higher efficiency potential compared to p-type Si. The combination of n-type Si with industrial low-cost techniques like screen-printing metallization enables the production of Si solar cells with high energy conversion efficiencies $\eta$ at low fabrication cost. For this reason, several n-type Si solar cell concepts have been developed in recent years such as n-PERT and n-type IBC solar cells. One application of n-PERT solar cells is using screen-printed metal contacts with H-grid design on both sides of the solar cells, resulting in a bifacial solar cell architecture. The bifaciality enables illumination of the solar cells from both sides and thereby capturing of light from the rear side which is scattered at the background. The power generation gain is in the range of 20%, strongly depending on the albedo of the background and mounting of the PV system, as reported in [130, 131].

However, it has been reported by several research groups that contact formation by screen-printing of metal pastes and subsequent contact firing causes a considerable loss of the open-circuit voltage $V_{oc}$ and thereby limits $\eta$ of the solar cells [132, 133]. It has been shown in [134, 135] that the use of pure Ag pastes of the older generation is insufficient in forming appropriate metal contacts on boron emitters in terms of a low specific contact resistance $\rho_c$. Note that current research indicates that using pure Ag pastes with optimized glass composition may result in appropriate metal contacts on boron emitters [136, 137]. However, usually Al containing Ag pastes are used for contact formation on boron emitters, since adding Al into Ag pastes reduces $\rho_c$ significantly, as shown in [134, 135, 138]. Although the addition of Al lowers $\rho_c$, it suffers from several disadvantages such as the formation of deep metal spikes, which can penetrate the SCR and thereby induce essential power loss because of enhanced recombination, as reported in [61, 135, 138-140]. Another drawback is the slightly higher line resistivity $r_l$ of Ag/Al metal fingers since Al has a lower conductivity than Ag, which causes slightly increased resistance related power losses. These metallization-induced power losses can overcompensate the benefits of using n-type Si. Thus, quantifying, analyzing and minimizing these power loss mechanisms is mandatory for increasing $\eta$ and lowering the LCoE.

Edler et al. showed in [76] that the metallization-induced $V_{oc}$ loss of bifacial n-PERT Si solar cells is around 30 – 40 mV for ideal contact firing conditions, which can mainly be associated with the Ag/Al metal contacts. However, the influence of different firing conditions on contact formation was not investigated. Since the firing conditions have a substantial impact, a detailed study on contact formation using screen-printing of Ag/Al pastes is presented in this chapter for a wide range of contact firing conditions. The focus is set on the formation of Ag/Al metal spikes and metallization-induced power losses of bifacial n-PERT solar cells with co-diffused boron emitters. The study on Ag/Al metal spike formation during contact firing is assisted by a detailed microstructural scanning electron microscopy (SEM) analysis.
3.1 Bifacial n-PERT solar cell process

In this section, the bifacial n-PERT Si solar cell process utilized within this work is presented. Figure 3.1 shows the process flow. As base material n-type Si substrates with $\rho_b = 6 – 9 \ \Omega cm$ are used. The wafers are alkaline textured on both sides. The alkaline texturing is conducted in an aqueous solution of KOH and an alcohol additive [141, 142] at 80°C. The anisotropic etching results in random pyramids on the Si surface reducing the surface reflectivity (see section 2.3.1). After texturing, the wafers are cleaned in a wet chemical piranha cleaning process (described in section 2.6.3). The thin oxide layer, which forms during the cleaning process, is removed in an aqueous solution of HF (2%). Thereafter, the SiO$_x$:B doping layer is deposited on the rear side of the wafers by means of PECVD. The doping layer must be protected from overcompensation by phosphorus atoms during sequential co-diffusion in a POCl$_3$ quartz tube furnace. Therefore, a SiN$_x$:H capping layer with a thickness of 120 nm is additionally deposited on top of the doping layer. The SiN$_x$:H capping layer can be deposited in the same PECVD device, subsequently to the deposition of the doping layer. The boron emitter and phosphorus FSF are formed in a sequential co-diffusion in a POCl$_3$ quartz tube furnace, according to the procedure outlined in section 2.5.2. Afterwards, the doping and capping layers as well as the PSG are removed in an aqueous solution of HF (10%). Formation of the s-FSF was realized by inkjet-printing of an etch-resistant ink and EB of the highly-doped FSF, corresponding to the procedure described in section 2.5.5. Then, the wafers are again cleaned in the piranha cleaning to remove potential contamination during formation of the s-FSF. The passivation of the s-FSF is established by means of thermal oxidation at 900°C for 5 minutes. This leads to a SiO$_2$ layer with a thickness of about 7 nm. Because the SiO$_2$ layer must be removed on the rear side prior to the deposition of the Al$_2$O$_3$, which is applied for the passivation of the boron emitter, an SiN$_x$:H layer with a thickness of 87 nm is deposited on the front side by PECVD. The thickness was calculated in accordance to the fact that during the removal of the SiO$_2$ layer on the rear side and the additional cleaning step, the SiN$_x$:H layer on the front side will also be etched. It was observed that $\approx 120$ s in an aqueous solution of HF (2%) are required for the complete removal of the SiO$_2$ layer. In the following piranha cleaning, another 120 s in an aqueous solution of HF (2%) is required for the removal of the thin oxide,
growing during chemical cleaning. With a measured etching rate of approximately 5 nm/min for the used SiN$_x$:H passivation layer in an aqueous solution of HF (2%), the post-process thickness can be estimated to be 67 nm. This matches the ideal thickness of the front ARC. Following this procedure, an Al$_2$O$_3$ layer with a thickness of 10 nm is deposited via PA-ALD, enabling excellent surface passivation on the boron emitter, as described in section 2.6.2. The Al$_2$O$_3$ layer is capped by a PECVD SiN$_x$:H layer with a thickness of 65 nm. Contact formation is established by screen-printing of a commercial Ag paste on the s-FSF and of a commercial Ag/Al paste on the boron emitter, as discussed in section 3.3. Contact firing is conducted in an IR belt conveyor furnace. Figure 3.2 depicts the schematic cross section of a bifacial n-PERT Si solar cell with co-diffused PECVD boron emitter.

![Schematic cross-section of a bifacial n-PERT solar cell](image)

**Figure 3.2:** Schematic cross-section of a bifacial n-PERT solar cell with co-diffused boron emitter and phosphorus FSF. Contact formation is conducted by screen-printing of a commercial Ag paste on the FSF and a commercial Ag/Al paste on the boron emitter. The emitter is passivated by an Al$_2$O$_3$/SiN$_x$ layer stack whereas the FSF is passivated by a SiO$_2$/SiN$_x$ layer stack.

### 3.2 Characterization of Si solar cells

The characterization of Si solar cells is important to determine the relevant IV-characteristics such as $\eta$, $j_{sc}$, $V_{oc}$, $FF$, $R_s$, $R_p$, $j_{01}$ and $j_{02}$ (see section 2.2). From these parameters conclusions can be drawn on the relevant loss mechanisms limiting $\eta$, which is mandatory to improve the solar cell fabrication processes. Therefore, the methods used in this work for characterizing Si solar cells are discussed in the following.

#### 3.2.1 IV-measurements

The measurement of the IV-curve of a Si solar cell under 1 sun illumination and in darkness allows the determination of the solar cell IV-parameters $\eta$, $j_{sc}$, $V_{oc}$ and $FF$. By fitting the two-diode model (see equation (2.8)) to the IV-curve, $j_{01}$, $j_{02}$, $R_s$ and $R_p$ can also be extracted. The measurement of the IV-curve under 1 sun illumination is executed in accordance with the IEC standard testing conditions (STC: incoming power density $p_{in} = 0.1$ W/cm$^2$, $T = 25^\circ$C and AM1.5g spectrum [17]). Commonly, a conductive metal chuck is used for the measurements of mono-facial solar cells with a full-area rear side metallization. The entire rear side of the solar cell is connected with the chuck. The front side metallization pattern is contacted with a setup...
of non-shading testing bridges with current and voltage pins, which are arranged above the BB of the solar cell. This setup corresponds well with the conditions in a mono-facial solar module and leads to reliable results for the measurement of mono-facial solar cells.

However, this setup leads to an overestimation of $\eta$ for IV-measurements of bifacial solar cells. First, $j_{sc}$ is increased due to transmitted long-wave light which is reflected at the reflective metal chuck and re-enters the solar cell bulk. The increase in $j_{sc}$ of around 0.4 - 0.5 mA/cm$^2$ depends on the solar cell structure and the metal chuck used, as discussed in section 3.4.5. Secondly, $R_S$ is underestimated and thereby $FF$ overestimated because the line resistance $R_L$ of the rear side metal fingers is unconsidered in that case. $R_L$ of the rear metal fingers must be considered for the application in bifacial solar modules because the solar cells are connected only at the BB on front and rear side. Thus, the current has to flow through the metal fingers before reaching the BB. In order to obtain reliable results also for bifacial solar cells, the IV-measurements were conducted in this work with a flasher from the manufacturer h.a.l.m, which works with testing bridges having current and voltage pins on both sides of the solar cell. The flasher is equipped with a non-reflective black back sheet.

### 3.2.2 Suns-Voc measurements

The principle of Suns-Voc measurements is to measure $V_{oc}$ in dependence on the incident light intensity during a slowly decaying flash [143-145]. A typical Suns-Voc measurement of a mono-facial n-PERT solar cell is depicted in Figure 3.3.

![Figure 3.3: Suns-Voc measurement of a mono-facial n-type PERT cell. In a) the light intensity and open-circuit voltage $V_{oc}$ are shown as a function of the measurement time during the decaying flash. In b) the pseudo light IV-curve and the pseudo power density are illustrated.](image)

The measurement is not affected by $R_S$ since it is performed under open-circuit conditions (no current is drawn from the cell). Only finished solar cells can be measured due to the fact that a p-n junction and metal contacts on both sides are required. The Suns-Voc measurement data provides information about material and passivation quality of the solar cell. In addition, the
IV-curve of the solar cell without influence of $R_S$ can be extracted if $j_{sc}$ is known. This allows the determination of the pseudo fill factor $pFF$, which describes $FF$ of the solar cell without influence of $R_S$. By comparing $FF_0$ and $pFF$, the influence of $R_{sh}$ and $j_{02}$ on $FF$ can be estimated. The comparison of $pFF$ and measured $FF$ allows to estimate the impact of $R_S$ on $FF$.

### 3.2.3 Transfer length method

Identifying and minimizing resistance related power losses induced by the metallization is an important issue for improving bifacial Si solar cells. The electrical resistance $R_c$, which can be associated with the metal-Si interface, contributes linearly to the total $R_S$ of a solar cell (see section 2.3.3). To optimize the contact formation, a reliable method for the measurement of the electrical quality of the metal-Si contacts is required. $R_c$ can directly be measured by means of transmission line, respectively transfer length method (TLM) [146-148]. In this method, two neighboring contacts with contact length $l_F$, width $w_F$ and distance $d$ between adjacent contacts are contacted by measurement probes, as illustrated in Figure 3.4. Each probe consists of a voltage and current pin, known as 4PP measurement.

The total resistance $R_T$ between neighboring contacts is measured by applying the voltage $V$ until a certain current $I$ flows between the contacts (commonly $I = 1$ mA) laterally through the conductive layer (e.g. boron emitter). $R_T$ can be calculated by the ratio of the applied voltage and current $V/I$. $R_T$ consists of the resistance of the metal contacts $R_M$, contact resistance $R_c$ and lateral resistance of the doped layer $R_D$ where the current flows through laterally

$$R_T = 2 \cdot R_M + 2 \cdot R_c + R_D.$$  \hspace{2cm} (3.1)

![Figure 3.4](image)

**Figure 3.4:** Schematic top-view (left side) and cross-section (right side) of the TLM pattern with increasing distance between neighboring fingers. The current flows laterally through the doped layer.

In general, $R_M$ can be neglected since it is much smaller than $R_D$ and $R_c$. $R_D$ depends on the sheet resistance $R_{sh}$ of the conductive layer, the distance $d$ between adjacent contacts and the width of the conductive layer which equals the contact length $l_F$. With this, the equation for $R_T$ can be simplified to

$$R_T(d) = 2 \cdot R_c + R_{sh} \cdot \frac{d}{l_F}. \hspace{2cm} (3.2)$$
Plotting $R_T$ as a function of $d$ allows the extrapolation of $R_{sh}$, $R_c$ and the transfer length $L_T$ by fitting a linear regression line (see Figure 3.5). $L_T$ is defined as the distance between the edge of the metal contacts, where most of the current flows into the metal contact, and the point where the current is reduced to 1/e of its initial value. This phenomenon is caused by a non-uniform current flow, resulting in current crowding at the edge of the local metal-Si contacts [147]. For contacts with high electrical quality in terms of low $R_c$, $L_T$ is smaller than $w_F$. The slope of the linear regression line equals $R_{sh}/l_F$ (see equation (3.2)), allowing to determine $R_{sh}$. $R_c$ can be calculated at the intersection of the linear regression line with the y-axis ($d = 0$) whereas $L_T$ can be calculated at the intersection of the linear regression line with the x-axis ($R_T = 0$), as shown in Figure 3.5.

![Figure 3.5: Plot of the total resistance $R_T$ as a function of the distance $d$ between adjacent contacts. The sheet resistance of the doped layer $R_{sh}$, contact resistance $R_c$ and transfer length $L_T$ can be quantified by fitting a linear regression line.](image)

Since the contact resistance $R_c$ is strongly dependent on the electrically active contact area $A_{c,el}$, the specific contact resistance $\rho_c$ can be calculated by

$$\rho_c = R_c \cdot A_{c,el},$$

which allows an area-independent quantification of the electrical quality of the contacts. Note that $A_{c,el}$ can be smaller than the physical contact area $A_c = l_F \cdot w_F$ for small $L_T$ due to the non-uniform carrier transport.

The theoretical description of $R_c$ can be deduced from the differential equations for the potential underneath the metal contacts assuming that the current flows only laterally in the doped layer and $l_F \gg w_F$, $L_T$ [149]

$$R_c = R_{sh} \cdot \frac{L_T}{l_F} \cdot \coth \left( \frac{w_F}{L_T} \right) = \frac{\rho_c}{l_F} \cdot L_T \cdot \coth \left( \frac{w_F}{L_T} \right).$$

(3.4)
Rewriting this equation, allows the calculation of the specific contact resistance by

\[ \rho_c = R_c \cdot l_F \cdot L_T \cdot \tanh \left( \frac{w_F}{L_T} \right). \] (3.5)

\( L_T \) can be calculated by [149]

\[ L_T = \sqrt{\frac{\rho_c}{R_{sh}}}. \] (3.6)

The measurement of \( \rho_c \) enables comparing the electrical quality of different contact geometries since it is area-normalized. Note that TLM measurements are only valid if the current flows in an infinitesimally thin conductive layer. Thus, TLM measurements can only be performed, if the polarity of the conductive layer is opposite to the polarity of the base material. Otherwise, the whole base contributes to lateral carrier transport, which causes an essential underestimation of the calculated values of \( R_{sh} \) and an overestimation of \( L_T \). For this reason, p-type base material is used for the TLM measurements of the Ag front contacts on the phosphorus-doped FSF. Measurement errors can also occur, if the metal contacts penetrate through the SCR, which results in local short-circuits between conductive layer and base. For the measurement of \( \rho_c \), a special TLM pattern (illustrated in Figure 3.6) was used within this work.

\[ \text{TLM: } w_F = 200 \, \mu m \quad d = 200, 400, ..., 2800 \, \mu m \]
\[ \text{measurement of } r_l: \quad w_F = 60, 70, ..., 120 \, \mu m \]

\[ \text{TLM: } w_F = 200 \, \mu m \quad d = 200, 400, ..., 2800 \, \mu m \]

**Figure 3.6:** TLM pattern for the measurement of the specific contact resistance \( \rho_c \) and line resistivity \( r_l \) for finger width of 60 – 120 \( \mu m \). The red lines indicate chip saw cuts where the emitter was electrically isolated.

The distance between neighboring fingers is varied from 200 - 2800 \( \mu m \) with a step size of 200 \( \mu m \). The width of the contacts is kept constant at \( w_F = 200 \, \mu m \). For reliable measurements of \( R_T \), it is required that the emitter underneath the TLM structure is electrically isolated. Therefore, lines were cut into the wafer surface by a chip saw to a depth of around 50 \( \mu m \) to ensure that the emitter was cut through. The cut lines are marked in Figure 3.6 as red lines. By this
procedure, the length of the contacts was set to \( l_F = 10 \) mm. In addition, the line resistivity \( r_l \) of the metal fingers for a width of \( 60 – 120 \mu \text{m} \) can be measured by a 4PP setup. Since it may be beneficial to use even narrower fingers, an adapted TLM pattern was designed, which enables the measurement of \( r_l \) in the range of \( 40 – 80 \mu \text{m} \) finger width.

### 3.3 Front side metallization

Screen-printing metallization is still limiting \( \eta \) of bifacial Si solar cells. Therefore, this section is focused on quantifying, analyzing and minimizing these metallization-induced power losses. As already briefly discussed in section 2.3, three different metallization-induced power loss mechanisms can be differentiated in general:

I. **Shading by the metal grid**  
   Power loss due to shading of the wafer surface by the metal grid are linearly correlated to the metal coverage \( f_c \). Therefore, the photo-generated current density \( j_{ph} \) reaches its maximum when \( f_c \) is zero. The computation of power loss due to shading is discussed in section 3.3.2.

II. **Resistance related power losses**  
The resistance related power losses are composed of the lateral resistances of the doped emitter \( R_e \) and FSF/BSF \( R_{FSF/BSF} \) between the metal contacts, contact resistance \( R_c \) at the metal-Si interfaces and line resistance \( R_l \) of the metal contacts on both sides of a bifacial Si solar cell (initially introduced in section 2.3.3). These resistance related power losses reach their minimum for full-area metallization (\( f_c = 100\% \)). Resistance related power losses are discussed more in detail in sections 3.3.2/3.3.3.

III. **Recombination power losses**  
Recombination losses induced by the metallization at the metal-Si interface and in the SCR, which can be described by an increased \( j_{01} \) and \( j_{02} \) according to the two-diode model (introduced in section 2.2), increase approximately linearly for an increase in \( f_c \). Power loss due to recombination induced by the metallization is discussed in more detail in section 3.3.4.

These power loss mechanisms show an opposing behavior for an increase in \( f_c \). Hence, there is a substantial tradeoff between shading and recombination losses on one side and resistance related losses on the other side. This tradeoff is discussed in more detail in the following by calculating the relevant power loss mechanisms based on measurement data (line resistivity \( r_l \), specific contact resistance \( \rho_c \) and saturation current density \( j_{01c} \) of screen-printed Ag metal contacts on the n⁺-doped FSF of bifacial n-PERT Si solar cells) and computing the ideal front side metallization pattern to minimize power losses induced by the metallization.
3.3.1 Screen-printing metallization

With a global market share of almost 98%, screen-printing was the prevailing technology for front side metallization of Si solar cells in 2016 [5]. It is predicted that screen-printing will continue its dominance in the PV market in the next decade due to its process simplicity and high throughput which is necessary for solar cell fabrication at low LCoE. Screen-printing can be applied to both sides of the solar cell, also enabling the fabrication of bifacial devices. Most frequently, a single screen-printing step is used to minimize process cost. The metal paste is pressed through local openings in a meshed screen by a squeegee and adheres to the wafer surface. The amount of the deposited paste and shape of the metal fingers depend strongly on the composition of the paste, the parameters of the mesh and screen-printing properties, as discussed in the following. A way to further improve the electrical quality of the screen-printed contacts and reduce power loss due to shading is the application of double printing (also known as print-on-print), where a second printing step is conducted to enhance the aspect ratio of the fingers and reduce possible interruptions [150]. Another approach for improving the electrical quality of screen-printed metal contacts is dual printing, where a non-contacting paste is used for the BB in order to decrease the recombination induced by the metal contacts [150, 151]. With state-of-the-art screen-printing technology, finger widths of 45 - 55 µm can be realized for industrial Si solar cell production by single printing and 40 - 50 µm by double printing using commercial Ag pastes [150]. On lab scale, the viability of 35 µm thick fingers was reported. Alternative methods for metallization of the front side are stencil printing [151, 152], dispensing [153], inkjet-printing [154] and light-induced plating [155]. In this work, screen-printing (single printing step) is exclusively used for the metallization of the front side.

Most frequently, Ag pastes are applied for the metallization of phosphorus-doped surfaces. The contact formation mechanism relies on Ag crystallites that grow into the Si substrate below the screen-printed contacts during contact firing, leading to contacts with high electrical quality in terms of low specific contact resistance \( \rho_c \), as reported e.g. in [156]. \( \rho_c \) is inversely related to the area density of the Ag crystallites. The metal contacts must additionally feature a high line conductivity, which is strongly dependent on the applied metal and shape of the metal contacts in terms of width and height, as well as an appropriate mechanical long-term adhesion. Despite its high market price of around 560 €/kg [5], Ag is deployed because of its high conductivity, which is the highest conductivity of all metals.

3.3.2 Finger width and line resistivity of front side metallization

Usually screen-printing a H-grid pattern is used for front side metallization. The H-grid pattern consists of \( N_F \) thin parallel fingers (finger width \( w_F \), effective finger length \( l_F \) and distance \( d_F \) between adjacent fingers) and perpendicularly to that \( N_{BB} \) busbars BB (width of the BB \( w_{BB} \) and length of the BB \( l_{BB} \)), as illustrated in Figure 3.7 for a 6-inch industrial Si solar cell (substrate length \( l_S \)). \( l_D \) is the distance between metallization and substrate edge, since the metallization is commonly not screen-printed completely to the substrate edge.
Figure 3.7: Schematic top view of a screen-printed H-grid pattern applied for typical 6-inch industrial Si solar cells with three busbars with the width $w_{BB}$, substrate length $l_s$, effective finger length $l_f$, finger distance $d_f$, finger width $w_f$ and the distance between metallization and edge of the substrate $l_d$.

One of the relevant parameters describing the electrical quality of screen-printed metal contacts is the line resistance $R_l$, which contributes linearly to $R_S$ (see equation (2.29)). $R_l$ depends on the width, height and shape of the screen-printed fingers. Since it is often not trivial to determine these values experimentally, commonly the line resistivity $r_l$ is measured by 4PP. $r_l$ includes all information about the mean cross section of the metal fingers. The shape of the fingers, and therefore $r_l$, depends strongly on the parameters of the mesh, metal paste composition, screen-printing parameters, such as printing velocity and the type of squeegee used for screen-printing. The mesh parameters are specified by the number of wires per inch and wires thickness. 400/18 specifies a mesh with 400 wires/inch with a wire thickness of 18 $\mu$m. The composition of the commercial Ag pastes used in this work is not known, since this is a corporate secret. However, it is assumed that the paste is composed of about 90% Ag powder, 5% glass frit and 5% organics [84]. The screen-printing parameters are the spacing between substrate and screen, contact pressure of the squeegee and printing velocity. The spacing between Si substrate and screen (1.8 mm) and squeegee pressure (1.2 bar) are kept constant during the whole study, whereas the printing velocity was varied between slow (50 mm/s) and fast (200 mm/s).

$r_l$ was determined by screen-printing a TLM pattern (introduced in section 3.2.3) with different mesh parameters and printing velocities, using the commercial Ag paste P1 and a common trailing squeegee, and subsequent 4PP measurement. The finger widths of the screen-printed Ag metal fingers, measured by optical microscopy, as a function of the screen opening are depicted in Figure 3.8 for different mesh parameters and printing velocities. Note that these presented values are averaged over at least three samples and the error bars correspond to the standard deviation. The width of the Ag metal fingers is always greater or equal to the screen opening and depends also strongly on the mesh parameters and wearing of the screen. The orange colored dashed line in Figure 3.8 represents the bisecting line, where the finger width equals the screen opening. Frequently used screens cause strong broadening of the fingers of around 25 – 40 $\mu$m. For new screens, the broadening of the fingers is only in the range of around
5 – 15 µm. It turned out that the printing speed (fast or slow) has only a negligible influence on the finger width. On the other hand, it is evident that the mesh parameters have a substantial influence on the finger width. Best results, in terms of narrow and homogeneous Ag metal fingers with only small broadening, could be achieved for mesh parameters 400/18, using a common trailing squeegee. It is assumed that the more distinct broadening of the fingers for frequently used screens is caused by abrasion of the emulsion over the mesh during cleaning of the screens for removing metal paste residues using clean room cloths and alcohol.

The relative power loss $P_S$ caused by shading of the front surface by the metal H-grid is linearly correlated to $f_c$, which equates to the ratio of the physical contact area $A_c$ and total cell area $A$

$$P_S = C_S \cdot f_c = C_S \cdot \frac{A_c}{A},$$  \hspace{1cm} (3.7)

with the constant $C_S \leq 1$ accounting for the shape of the metallization. $C_S$ equals 1 in the case of rectangular shaped fingers and BB and is < 1 if part of the at the metallization reflected light can still be absorbed by the solar cell, e.g. if the metal finger has rounded edges. The following calculations are based on homogeneously shaped metal fingers and BB and can be modified for tapered fingers and BB easily. The metallization fraction $f_c$ includes the physical metallization area of the fingers $A_{c,F}$ and BB $A_{c,BB}$

$$f_c = \frac{A_c}{A} = \frac{A_{c,F} + A_{c,BB}}{A} = \frac{\left(2 \cdot N_{BB} \cdot l_F \cdot w_F \cdot N_F + (N_{BB} \cdot l_{BB} \cdot w_{BB}) \right)}{l_S^2}.$$  \hspace{1cm} (3.8)
$N_F$ can be calculated by

$$N_F = \frac{l_S - 2 \cdot l_D}{d_F} + 1. \quad (3.9)$$

Inserting equation (3.9) into equation (3.8) allows the calculation of $P_S$ as a function of $w_F, d_F$

$$P_S(w_F, d_F) = C_S \cdot \left( \frac{2 \cdot N_{BB} \cdot l_F \cdot w_F \cdot \left( \frac{l_S - 2 \cdot l_D}{d_F} + 1 \right) + (N_{BB} \cdot l_{BB} \cdot w_{BB})}{l_S^2} \right) \approx C_S \cdot \left( \frac{w_F}{d_F} + \frac{N_{BB} \cdot w_{BB}}{l_S} \right). \quad (3.10)$$

For the simplification, $l_D$ and the 1 are neglected and it is assumed that $2 \cdot N_{BB} \cdot l_F$ and $l_{BB}$ approximate $l_S$. For commonly used H-grid metallization ($w_F < 120 \mu m$, $1.5 \, mm < d_F < 2.5 \, mm$ and $w_{BB} = 1.2 \, mm$), the relative power loss due to shading is less than 10%.

By the transition to narrow fingers and/or high finger spacing $d_F$ in order to minimize power loss caused by shading and recombination (reviewed in section 3.3.4), resistance related power losses can become limiting for $FF$ of the solar cells mainly due to a high line resistivity $r_l$ and/or lateral resistance in the FSF $R_{FSF}$, as discussed in the following. Thus, screen-printing narrow Ag metal fingers is not the only requirement for an appropriate front side metallization.

The metal fingers must also have a sufficiently low $r_l$ to ensure low series resistance $R_S$ and high $FF$ of the bi-facial n-PERT solar cells (see section 2.3.4). The measured $r_l$ of commercial Ag paste P1, as a function of the measured finger width for different mesh parameters and printing velocities, is depicted in Figure 3.9 using a common trailing squeegee. Note that the presented values are averaged over at least three samples and the error bars correspond to the standard deviation. As expected, $r_l$ strongly increases for decreasing finger width ($r_l \sim 1/w_F$). Moreover, the possibility of narrowing and interruption of metal fingers becomes considerable for narrow fingers, which results in enhanced $R_S$ and thereby increased resistance related power losses. A typical interruption of an Ag finger is illustrated in Figure 3.10.

Although the printing velocity has only a minor influence on $w_F$ (see Figure 3.8), it is observed that the printing velocity significantly impacts $r_l$ of narrow fingers, especially for the mesh with the parameters 400/18 (compare filled and blank red triangles in Figure 3.8). Altogether, it turned out that best results, in terms of homogeneous narrow fingers with low $r_l$, can be achieved for the mesh with parameters 400/18 at low printing velocity of $50 \, mm/s$, using commercial Ag paste P1 and a common trailing squeegee. Moreover, it was observed that $r_l$ is independent of the set peak firing temperatures $T_S$ in the range of $800 – 890^\circ C$ at a belt speed of $6 \, m/min$ (data not shown). Note that the wafer temperature $T_W$ is essentially lower than $T_S$, as discussed in section 3.4.4.

In the following the mathematical expressions for the calculation of the power loss due to the line resistivity $r_l$ and lateral resistance in the s-FSF $R_{FSF}$ as a function of $w_F$ and $d_F$ are derived.
Metallization-Induced Power Loss of Bifacial n-Type Silicon Solar Cells

Figure 3.9: Line resistivity $r_l$ as a function of the finger width, measured by optical microscopy, for different mesh parameters and printing velocities using a common trailing squeegee and commercial Ag paste P1. Note that the lines are only guides to the eye.

![Figure 3.9](image)

Figure 3.10: Light-optical micrograph of an interruption of a narrow (width ≈ 60 µm) screen-printed Ag metal finger. Interruptions lead to substantially high line resistivities $r_l$ and have to be avoided.

![Figure 3.10](image)

The current flow through a finger element $dx$ at distance $x$ can be calculated according to the unit cell (see Figure 3.11) by [12]

$$I(x) = j_{MPP} \cdot (d_F - w_F) \cdot x,$$

assuming a homogeneous charge carrier generation between the metal fingers. The incremental resistance of the element $dx$ is

$$dR_l = \rho \cdot \frac{dx}{w_F \cdot h_F} = r_l(w_F, h_F) \cdot dx,$$

with the resistivity of the metal $\rho$ and finger height $h_F$. Since the reliable measurement of $h_F$ is not trivial, the line resistivity $r_l$ is used, which also contains information on $h_F$. 

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The incremental power loss in the section $dx$ of the finger is given by
\[ dP_l = I(x)^2 \cdot dR_l = j_{MPP}^2 \cdot (d_F-w_F)^2 \cdot x^2 \cdot r_l(w_F, h_F) \cdot dx. \] (3.13)

The total power loss $P_{l,\text{tot}}$ due to $r_l$ of the metal finger can be calculated by
\[ P_{l,\text{tot}}(w_F, d_F) = \int_0^{l_F} dP_l = \int_0^{l_F} j_{MPP}^2 \cdot (d_F-w_F)^2 \cdot r_l(w_F, h_F) \cdot x^2 \cdot dx \] (3.14)
\[ = j_{MPP}^2 \cdot (d_F-w_F)^2 \cdot r_l(w_F, h_F) \cdot l_F^3/3. \]

The maximum power generated in the unit cell (see Figure 3.11) at MPP is
\[ P_{l,\text{max}}(d_F) = I_{MPP} \cdot V_{MPP} = j_{MPP} \cdot l_F \cdot (d_F-w_F) \cdot V_{MPP}. \] (3.15)

The relative power loss in the fingers $P_l$ can be calculated by the fraction of $P_{l,\text{tot}}$ and $P_{l,\text{max}}$
\[ P_l(w_F, d_F) = \frac{P_{l,\text{tot}}}{P_{l,\text{max}}} = \frac{j_{MPP}^2 \cdot (d_F-w_F)^2 \cdot r_l(w_F, h_F) \cdot l_F^3/3}{j_{MPP} \cdot l_F \cdot (d_F-w_F) \cdot V_{MPP}} \] (3.16)
\[ = \frac{r_l(w_F, h_F)}{3} \cdot (d_F-w_F) \cdot l_F^2 \cdot \frac{j_{MPP}}{V_{MPP}}. \]

The area-normalized series resistance attributed to the line resistance of the fingers $R_l$ is
\[ R_l = \frac{P_{l,\text{tot}}}{I^2} \cdot A_U = \frac{j_{MPP}^2 \cdot (d_F-w_F)^2 \cdot r_l(w_F, h_F) \cdot l_F^3/3}{(j_{MPP} \cdot l_F \cdot (d_F-w_F))^2} \cdot d_F \cdot \left( l_F + \frac{w_{BB}}{2} \right) \] (3.17)
\[ = \frac{r_l(w_F, h_F)}{3} \cdot l_F \cdot d_F \cdot \left( l_F + \frac{w_{BB}}{2} \right) \approx \frac{r_l(w_F, h_F)}{3} \cdot l_F \cdot d_F^2, \]

with the area of the unit cell $A_U$. Note that $R_l$ and $P_l$ are proportional to $r_l$, the square of $l_F$ and in addition proportional to $d_F$, as the current flowing through the finger depends linearly on $d_F$.

**Figure 3.11:** Schematic of the unit cell for deriving a mathematical expression to compute the power loss by line resistance of the metal fingers and lateral resistance of the s-FSF between the metal fingers.
The lateral resistance of the highly doped region such as s-FSF $R_{FSF}$ (or selective emitter $R_e$) depends on the sheet resistance of the lowly doped area $R_{sh,l}$, sheet resistance of the highly doped area $R_{sh,h}$, $w_F$ and $d_F$. With the assumptions that the BB do not contribute to collection of charge carriers and homogeneous generation of charge carriers in the highly and lowly doped regions between the Ag metal fingers, the photo-generated current at MPP increases linearly between the fingers as depicted in Figure 3.11

$$I(y) = j_{MPP} \cdot l_F \cdot y.$$  \hfill (3.18)

The incremental power loss in the section $dy$ of the FSF (or emitter) is given by

$$dP_{FSF} = I(y)^2 \cdot dR_{FSF} = I(y)^2 \cdot R_{sh} \cdot dy/l_F.$$  \hfill (3.19)

The total power loss in the s-FSF (or emitter) due to lateral charge carrier transport can be calculated by integrating the incremental power loss considering the highly doped region with width $w_h$ and the lowly doped region with width $w_l$ (see Figure 3.11)

$$P_{FSF,tot} = \int_0^{w_h} j_{MPP}^2 \cdot l_F^2 \cdot R_{sh,h} / l_F \cdot y^2 \cdot dy + \int_{w_l}^{(d_F-w_F)/2} j_{MPP}^2 \cdot l_F^2 \cdot R_{sh,h} / l_F \cdot y^2 \cdot dy.$$  \hfill (3.20)

The upper bound of the second integration stems from the fact that the current flows primarily at the edge into the metal contacts for low $\rho_c$ (see section 3.2.3). Solving the integrals gives

$$P_{FSF,tot}(w_F, d_F) = j_{MPP}^2 \cdot l_F^2 \cdot (R_{sh,h} \cdot (d_F-w_F)^3/8 + \Delta R_{sh} \cdot w_l^3)/3,$$  \hfill (3.21)

with $\Delta R_{sh} = R_{sh,l} - R_{sh,h}$. The maximum power generated in the unit cell (the smallest unit cell used for this calculation is the lower half of the unit cell shown in Figure 3.11) at MPP is

$$P_{FSF,max}(w_F, d_F) = l_{MPP} \cdot V_{MPP} = j_{MPP} \cdot \left(l_F + \frac{w_{BB}}{2}\right) \cdot (d_F-w_F)/2 \cdot V_{MPP}.$$  \hfill (3.22)

The relative power loss in the FSF $P_{FSF}$ can be calculated by

$$P_{FSF}(w_F, d_F) = \frac{P_{FSF,tot}}{P_{FSF,max}} = \frac{j_{MPP}^2 \cdot l_F^2 \cdot (R_{sh,h} / 8 \cdot (d_F-w_F)^3 + \Delta R_{sh} \cdot w_l^3)}{3 \cdot j_{MPP} \cdot \left(l_F + \frac{w_{BB}}{2}\right) \cdot \frac{(d_F-w_F)^2}{2} \cdot V_{MPP}}$$  \hfill (3.23)

$$= \frac{l_F}{l_F + \frac{w_{BB}}{2}} \cdot \left(R_{sh,h} \cdot \frac{(d_F-w_F)^2}{12} + \frac{2 \cdot \Delta R_{sh} \cdot w_l^3}{3 \cdot (d_F-w_F)}\right) \cdot \frac{j_{MPP}}{V_{MPP}}.$$

The simplification relies on the fact that $w_F$ is much smaller than $d_F$ and $w_{BB}$ smaller than $l_F$. Though, this simplification leads to substantial relative errors ($> 10\%$) for low $d_F$. Since $P_{FSF}$ is proportional to the square of $d_F$, it becomes the main contribution to the total power loss of a solar cell for wider finger spacing (large $d_F$), as shown in the following. The area-normalized
series resistance attributed to the emitter $R_{FSF}$ can be derived by the total power loss $P_{FSF,tot}$ divided by the square of the current which is generated in the unit cell multiplied by the area of the unit cell $A_U$

\[
R_{FSF}(w_F,d_F) = \frac{P_{FSF,tot}}{I^2} \cdot A_U
\]

\[
= \frac{j_{mpp}^2 \cdot l_F \cdot (R_{sh,h}(d_F-w_F)^3/8 + \Delta R_{sh}w_i^3)}{3 \cdot \left(\frac{j_{mpp} \cdot l_F \cdot (d_F-w_F)}{2}\right)^2} \cdot \frac{d_F}{2} \cdot \left(1 + \frac{w_{BB}}{2 \cdot l_F}\right)
\]

\[
= \left(\frac{R_{sh,h} \cdot (d_F-w_F)}{12} + \frac{2 \cdot \Delta R_{sh}}{3} \cdot \frac{w_i^3}{(d_F-w_F)^2}\right) \cdot \frac{d_F}{2} \cdot \left(1 + \frac{w_{BB}}{2 \cdot l_F}\right)
\]

\[
\approx \left(\frac{R_{sh,h} \cdot d_F^2}{12} + \frac{2 \cdot \Delta R_{sh}}{3} \cdot \frac{w_i^3}{d_F}\right).
\]

For the simplification, it is assumed that $w_F << d_F$ and $w_{BB} << l_F$. However, this simplification leads to substantial relative errors (> 10%) for low $d_F$, likewise to equation (3.23). The absolute values and behavior of $P_S$, $P_I$, and $P_{FSF}$ for a variation in $w_F$, $d_F$ and their contribution to the total power loss of n-PERT solar cells are discussed in the following.

### 3.3.3 Contact resistance of front side metallization

Another relevant parameter (besides $\rho_i$) describing the electrical quality of metal contacts is the area-normalized specific contact resistance $\rho_c$. As outlined in section 3.2.3, the TLM technique enables the measurement of $\rho_c$. TLM measurements are based on screen-printing a special TLM pattern (see Figure 3.6) and 4PP measurement of the total resistance $R_T$ between neighboring fingers. From this experimental data, the sheet resistance $R_{sh}$ of the conductive layer, $\rho_c$ and the transfer length $L_T$ can be extracted, according to equation (3.2). Since the current must flow laterally through an infinitesimally thin layer between the metal contacts, p-type Si substrates must be used instead of n-type Si substrates for the TLM measurements of Ag metal contacts on the n+-doped FSF. No measurable difference in the n+ doping profile by means of PCl$_3$ diffusion was observed between using n- or p-type Si substrates. Initially, the utilized p-type monocrystalline Si substrates were alkaline textured and cleaned prior to the PCl$_3$ diffusion. Subsequently, the PSG was removed using an aqueous solution of HF (2%) and one part of the wafers was etched back to around 100 $\Omega$/sq, according to the process sequence described in section 2.5.5. Finally, a SiN$_x$:H layer with a thickness of about 75 nm was deposited via PECVD before screen-printing and contact firing in an IR belt furnace for set peak firing temperatures $T_S$ in the range of 800 – 890°C. Note that the wafer peak temperature $T_W$ is essentially lower than $T_S$, as discussed in section 3.4.1.

$R_c$ at the metal-Si interface depends strongly on the composition of the Ag paste, n+ doping profile, surface morphology, passivation layer and $T_S$. As discussed in section 2.5.5, the n+-doped FSF is selectively etched-back from initially 30 – 40 $\Omega$/sq to around 100 $\Omega$/sq. $\rho_c$ was
measured on the highly-doped areas as well as on the etched-back areas to also obtain values for the worst case when screen-printing the Ag paste in the etched-back areas, caused by insufficient alignment. The results are depicted in Figure 3.12a for a commercial Ag paste of the older generation P3. The \( \rho_c \) values on the etched-back area \((R_{sh} \approx 100 - 130 \ \Omega/sq)\) depend strongly on \( T_S \) and may exhibit rigorous values for \( T_S < 860^\circ C \). \( \rho_c \) on the highly-doped area \((R_{sh} \approx 44 \ \Omega/sq)\) show very low values below 1 m\( \Omega \)cm\(^2\) for the whole range of \( T_S \), shown in Figure 3.12a for the commercial Ag paste P3 of an older generation and in Figure 3.12b for commercial pastes of a newer generation P1/P2. Note, that P1 was used for the studies of the line resistivity presented in section 3.3.1 and was exclusively used for metallization of the bifacial n-PERT Si solar cells. The results confirm the excellent contact formation of commercial Ag pastes on the highly n\(^+\)-doped FSF, even for pastes of the older generation, and the necessity of good alignment during screen-printing.

![Figure 3.12](image)

**Figure 3.12:** Specific contact resistance \( \rho_c \) on a phosphorus-doped FSF formed by POCl\(_3\) diffusion. In a) the values for a commercial Ag paste of an older generation are depicted for the highly-doped areas (44 \( \Omega/sq\)) underneath the metal contacts as well as the etched-back areas (100 - 130 \( \Omega/sq\)). In b) the values for commercial Ag pastes of a newer generation are illustrated for the highly-doped areas (40 \( \Omega/sq\)) of the FSF. The values are averaged over at least three samples and the error bars correspond to the standard deviation. Note that the lines are only guides to the eye.

The area-normalized contact resistance of front/back metal contacts \( R_{c,f}/R_{c,b} \), quoted as \( R_c \) in the following, can be calculated with the area-normalized specific contact resistance of the metal fingers \( \rho_c \)

\[
R_c = \rho_c \cdot \frac{A}{A_{c,F}} = \rho_c \cdot \frac{l_s^2}{2 \cdot N_{BB} \cdot l_F \cdot w_F \cdot \left(\frac{l_s - 2 \cdot l_d}{d_F} + 1\right)} \approx \rho_c \cdot \frac{d_F}{w_F},
\]

for a H-grid according to Figure 3.7, assuming that only the fingers contribute to the collection of charge carriers. This is especially true for dual-printing technology, where a non-contacting paste is used for the BB and leads only to small errors for contacted BB. For the simplification
of equation (3.25), the same assumptions were made as used for simplifying equation (3.8). The determination of \( \rho_c \) by TLM is introduced in section 3.2.3. Note that the non-contacted area between the end of the metal fingers and the edge of the solar cell is also neglected in the simplified form of equation (3.25).

The relative power loss \( P_c \) at MPP due to \( R_c \) at the metal-Si interface can be estimated with \( P = I^2 R \) to be

\[
P_c = \frac{I_{MPP}^2 \cdot R_c/A}{I_{MPP} \cdot V_{MPP}} = \frac{I_{MPP} \cdot R_c}{V_{MPP} \cdot A} = R_c \cdot \frac{j_{MPP}}{V_{MPP}}.\tag{3.26}
\]

Inserting equation (3.25) into equation (3.26) allows the determination of \( P_c \) as a function of \( w_F \) and \( d_F \)

\[
P_c(w_F, d_F) = \rho_c \cdot \frac{A}{2 \cdot N_{BB} \cdot l_F \cdot w_F \cdot \left(\frac{l_S - 2 \cdot l_D}{d_F} + 1\right)} \cdot \frac{j_{MPP}}{V_{MPP}} \approx \rho_c \cdot \frac{d_F \cdot j_{MPP}}{w_F \cdot V_{MPP}}.\tag{3.27}
\]

Based on the above-reviewed power loss mechanisms, their absolute values and behavior for a variation in \( w_F \) and \( d_F \) was calculated to get a better understanding of these loss mechanisms and to quantify their impact on the total power loss of n-PERT solar cells. The line resistivity \( r_l(w_F, d_F) \) was determined by fitting the function \( f(w_F) = a/w_F^b \) to the measured \( r_l \) values for the mesh parameters 400/18 using low printing velocity (see Figure 3.13). The exponent of \( w_F \) in the fit function is larger than one, which means that the effective height of the Ag metal fingers also depends strongly on \( w_F \) and decreases for decreasing \( w_F \). This corresponds well to the observations by optical microscopy.

**Figure 3.13:** Fitting the measured line resistivity \( r_l \) as a function of the finger width for the mesh 400/18, used for the fabrication of bifacial n-PERT solar cells in this work, using low screen printing velocity. Note that the lines are only guides to the eye.
Figure 3.14 illustrates the computed front side metallization-induced relative power losses by shading $P_S$ (equation (3.10)), line resistance of the metal fingers $P_l$ (equation (3.16)), lateral resistance of the phosphorus FSF/emitter $P_{FSF}/P_e$ (equation (3.23)), contact resistance of the Ag metal fingers $P_c$ (equation (3.27)) and the total power loss $P_{total} = P_S + P_c + P_{FSF}/P_e + P_l$ as a function of finger spacing $d_F$ ($w_F = 70 \mu m$) for a) a 6-inch RJ n-PERT Si solar cell with 40/100 $\Omega$/sq n$^+$-doped s-FSF and b) a 6-inch p-type Si solar cell with 40/100 $\Omega$/sq selective n$^+$-doped emitter. Furthermore, these relative power losses are illustrated in Figure 3.14c/d as a function of finger width $w_F$ ($d_F = 2 \text{ mm}$) for c) a 6-inch RJ n-PERT Si solar cell and d) a 6-inch p-type Si solar cell.

![Figure 3.14: Power loss mechanisms related to the front side metallization as a function of the finger spacing $d_F$: a) for a 6-inch n-type RJ PERT Si solar cell with 40/100 $\Omega$/sq selective n$^+$-doped FSF and b) for a 6-inch p-type Si solar cell with 40/100 $\Omega$/sq selective n$^+$-doped emitter, as well as a function of finger width $w_F$: c) for a 6-inch n-type RJ PERT Si solar cell and d) for a 6-inch p-type Si solar cell. The lines are only guides to the eye.](image_url)

The experimental data used for the calculation of the power losses is presented in Table 2. Since the whole n-type base contributes to lateral carrier transport between the fingers for the n-PERT cell, $P_{FSF}$ is about 1% absolute lower than $P_e$ for the solar cell with p-type base for ideal finger spacing. This is one of the benefits of applying RJ n-PERT Si solar cells. For the screen-printed
Ag metal fingers with $w_F = 70 \, \mu m$ and measured specific contact resistance $\rho_c = 0.5 \, m\Omega cm^2$, the power loss due to contact resistance can be neglected. The narrow Ag fingers with measured $r_l = 33 \, m\Omega/mm$ have a noticeable contribution to $P_{total}$. The optimal finger spacing is mainly determined by a tradeoff between power loss due to shading and lateral resistance between the fingers and can be calculated to be $d_F = 2.03 \, mm$ for the 6-inch n-PERT cells and $d_F = 1.75 \, mm$ for the 6-inch p-type Si solar cells ($w_F = 70 \, \mu m$). Similar results are obtained for the additionally processed 5-inch n-PERT solar cells, leading to an ideal spacing $d_F = 1.93 \, mm$ for the 5-inch n-PERT cells (data not shown). This slightly lower ideal finger spacing for minimal power loss relies on the fact that the effective finger length $l_F$ is larger (compare values for $l_F$ in Table 2) because only two BB are applied for the 5-inch solar cells, instead of three BB for the 6-inch cells. This leads to an increased contribution of the line resistance of the Ag metal fingers, which is proportional to the square of $l_F$ (equation (3.17)).

Table 2: Used parameters for the computation of the power losses induced by the front metallization based on experimentally measured data.

<table>
<thead>
<tr>
<th>Cell size</th>
<th>$C_S$</th>
<th>$l_F$</th>
<th>$N_{BB}$</th>
<th>$l_{BB}$</th>
<th>$w_{BB}$</th>
<th>$l_S$</th>
<th>$l_D$</th>
<th>$\rho_c$</th>
<th>$r_l$</th>
<th>$j_{MPP}$</th>
<th>$V_{MPP}$</th>
</tr>
</thead>
<tbody>
<tr>
<td>6-inch</td>
<td>0.95</td>
<td>25.37</td>
<td>3</td>
<td>154</td>
<td>0.6</td>
<td>156</td>
<td>1.0</td>
<td>0.5</td>
<td>30,922/$w_F^{1.61}$</td>
<td>35.0</td>
<td>560</td>
</tr>
<tr>
<td>5-inch</td>
<td>0.95</td>
<td>30.45</td>
<td>2</td>
<td>123</td>
<td>0.6</td>
<td>125</td>
<td>1.0</td>
<td>0.5</td>
<td>30,922/$w_F^{1.61}$</td>
<td>35.0</td>
<td>560</td>
</tr>
</tbody>
</table>

Power loss caused by recombination at the metal contacts is not included in the computation depicted in Figure 3.14, since it is not trivial to derive an equation for the relative power loss induced by recombination. However, power loss due to recombination at the metal contacts must be considered, especially when applying metal contacts with a high saturation current density $j_{0C}$. To solve this issue, the optimal H-grid is computed in the following by implementing the different loss mechanisms into the two-diode model. This allows also to account for recombination at the metal contacts by implementing their saturation current densities, as discussed in the following.

### 3.3.4 Saturation current density of front side metallization

Besides shading and resistance related loss mechanisms, recombination power losses at the metal contacts must also be considered for optimizing the H-grid metallization. Recombination can be described by the saturation current densities of the first and second diode $j_{01}/j_{02}$ in the two-diode model (see equation (2.8) in section 2.2). Applying a SiO$_2$/SiN$_x$:H passivation layer stack (see section 2.6.1) ensures a low surface recombination velocity of the phosphorus FSF, resulting in a low $j_{01}$ of around 50 fA/cm$^2$ in the etched-back areas ($R_{sh} \approx 100 \, \Omega/sq$) and of around 200 fA/cm$^2$ in the non-etched areas ($R_{sh} \approx 40 \, \Omega/sq$) for alkaline textured surfaces, as published by Book et al. in [97], using the same passivation and EB processes as used in this
work. To determine the total saturation current density of the front surface \( j_{01,f} \), it is required to also consider the saturation current density of the metallized part \( j_{01c} \). The passivation layer is almost entirely etched by the glass frit in the Ag metal paste during contact firing for forming a direct metal-Si contact by Ag crystallites which grow into the Si surface [158]. The contact formation mechanism is strongly dependent on the metal paste composition, crystal orientation and contact firing conditions, as reported in [156]. It is assumed that all charge carriers reaching the metal contacts recombine. Hence the surface recombination velocity in the contacted area \( S_c \) equals the thermal limit of charge carrier transport \( (S_c = 10^7 \, \text{cm/s}). \) \( j_{01c} \) is strongly dependent on electrical shielding of the Ag contacts by the FSF and decreases, in contrary to the passivated area \( (S_{pass} << S_c) \), with decreasing \( R_{sh} \) of the FSF. Therefore, a highly-doped phosphorus FSF underneath the Ag metal contacts is beneficial in terms of electrical shielding and to ensure a low contact resistance (see section 3.3.3). For screen-printed Ag metal contacts on the non-etched phosphorus doping profile \( (R_{sh} \approx 40 \, \Omega/\text{sq}) \), \( j_{01c} \) increases from 400 to 600 fA/cm\(^2\) for an increase of \( T_s \) from 840 to 920°C, as reported in [97]. The saturation current densities of the etched (lowly doped) \( j_{01,f} \) and non-etched (highly doped) \( j_{01,h,f} \) non-contacted (passivated) areas of the FSF, as well as \( j_{01c} \) contribute to \( j_{01f} \) according to

\[
j_{01f} = f_l \cdot j_{01,f} + f_h \cdot j_{01,h,f} + f_c \cdot j_{01c} \tag{3.28}
\]

with the fractions of the lowly and highly doped areas \( f_l \) and \( f_h \), as well as the metallization fraction \( f_c \). Since the non-etched (masked) area is larger than the metallized area due to alignment issues, \( f_h \) can be calculated by subtracting \( f_c \) from the fraction \( f_{mask} \) \( (f_h = f_{mask} - f_c) \). \( f_l \) can be calculated by \( f_l = 1 - f_{mask} \). For a typical 6-inch 3BB H-grid with \( w_F = 70 \, \mu\text{m}, \ d_F = 2 \, \text{mm}, \ w_{BB} = 600 \, \mu\text{m} \) \( (f_c \approx 4.6\%) \), finger width \( w_{M,F} = 250 \, \mu\text{m} \) and BB width \( w_{M,BB} = 800 \, \mu\text{m} \) of the etch-resistant mask \( (f_{mask} \approx 13.9\%) \) and the saturation current densities of the different doped areas outlined above, \( j_{01f} \) can be estimated to be around 85 fA/cm\(^2\) for ideal firing conditions.

The metallization can also cause an increased \( j_{02} \) if the SCR is affected by the contact formation. Analogous to equation (3.28), the total saturation current density of the second diode of the front surface \( j_{02f} \) can be calculated by

\[
j_{02f} = f_l \cdot j_{02,f} + f_h \cdot j_{02,h,f} + f_c \cdot j_{02c} \tag{3.29}
\]

With \( j_{02c} \approx 20 \, \text{nA/cm}^2 \) of the Ag metal contacts on the non-etched phosphorus doping profile \( (R_{sh} \approx 40 \, \Omega/\text{sq}) \), reported by Fellmeth et al. in [86], and assuming \( j_{02,f} = j_{02,h,f} = 5 \, \text{nA/cm}^2 \), \( j_{02f} \) can be estimated to be around 5.7 nA/cm\(^2\) for ideal firing conditions.

These low \( j_{01f} \) and \( j_{02f} \) values indicate that the power loss due to recombination induced by the Ag front side metallization is small compared to the power loss by shading and series resistance (the influence of \( j_{01} \) and \( j_{02} \) on the IV-characteristics is reviewed in section 2.3.4). With the measured values of the finger width \( w_F \), line resistivity \( r_l \), contact resistance \( \rho_c \) and saturation current density \( j_{0c} \) of screen-printed Ag metal fingers, the power loss mechanisms related to the Ag front side metallization are calculated in the following.
3.3.5 Optimization of front side metallization grid

To account for all relevant metallization-induced power loss mechanisms (shading, resistance and recombination) and to optimize the front side metallization H-grid of n-PERT solar cells, the whole IV-curves were computed using the two-diode model (see equation (2.8)). The IV-characteristics $j_{sc}, V_{oc}, FF$ and $\eta$ were then extracted from the IV-curves. The implemented total series resistance $R_s(w_F, d_F)$ comprises the contact resistance of the Ag-Si interface $R_c(w_F, d_F)$ (equation (3.25)), lateral resistance between the Ag metal fingers in the s-FSF $R_{FSF}(w_F, d_F)$ (equation (3.24)) and line resistance of the Ag metal fingers $R_l(w_F, d_F)$ (equation (3.17)). Note that the non-simplified equations were used. The line resistivity $r_l(w_F, d_F)$ was implemented by the fit function $f(w_F) = 30,922/w_F^{1.61}$ (see Figure 3.13). All parameters of the rear side (contact resistance $R_{c,b} = 0.05 \text{ } \Omega \text{cm}^2$, lateral resistance of the emitter $R_e = 0.1 \text{ } \Omega \text{cm}^2$, line resistance $R_{l,b} = 0.1 \text{ } \Omega \text{cm}^2$ and recombination at the rear side) were kept constant in the computation, since they do not affect the characteristics of the front side metallization.

Since shading causes a loss of the photo-generated current density $j_{ph}$, linearly correlated to the metallization fraction $f_c$, it was implemented according to

$$j_{ph}(f_c) = j_{ph,max} \cdot (1 - C_s \cdot f_c),$$

with $j_{ph,max} = 40.8 \text{ } \text{mA/cm}^2$ for alkaline textured wafers with SiO$_2$/SiN$_x$:H passivation. Power loss due to recombination at the front side Ag metallization was implemented according to equation (3.28), using the saturation current densities $j_{01}$ and $j_{02}$ reviewed in section 3.3.4. The fixed values used for the computation of the IV-curves of the bifacial 6-inch RJ n-PERT Si solar cells with phosphorus-doped s-FSF for different front side metallization parameters are presented in Table 3.

<table>
<thead>
<tr>
<th>$l_s$ [mm]</th>
<th>$l_d$ [mm]</th>
<th>$l_f$ [mm]</th>
<th>$N_{BB}$</th>
<th>$w_{BB}$ [mm]</th>
<th>$R_e$ [Ωcm$^2$]</th>
<th>$R_{c,b}$ [Ωcm$^2$]</th>
<th>$R_{l,b}$ [Ωcm$^2$]</th>
</tr>
</thead>
<tbody>
<tr>
<td>156</td>
<td>1.0</td>
<td>25.37</td>
<td>3</td>
<td>0.6</td>
<td>0.1</td>
<td>0.05</td>
<td>0.1</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>$\rho_c$ [mΩcm$^2$]</th>
<th>$r_l$ [mΩ/mm]</th>
<th>$j_{01,c}$ [fA/cm$^2$]</th>
<th>$j_{01,l}$ [fA/cm$^2$]</th>
<th>$j_{01,h,f}$ [fA/cm$^2$]</th>
<th>$j_{02,c}$ [nA/cm$^2$]</th>
<th>$w_{BB}$ [μm]</th>
<th>$w_{FF}$ [μm]</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.5</td>
<td>30,922/$w_F^{1.61}$</td>
<td>500</td>
<td>50</td>
<td>200</td>
<td>20</td>
<td>800</td>
<td>250</td>
</tr>
</tbody>
</table>

Figure 3.15 shows the IV-characteristics, extracted from the computed IV-curves of the solar cells, for a wide range of $w_F$ (10 - 140 μm) and $d_F$ (1 - 3 mm). Figure 3.15a illustrates that an increase in $f_c$ by increasing $w_F$ and/or decreasing $d_F$ causes a substantial power loss caused by shading, which lowers $j_{sc}$. On the other hand, by increasing $f_c$, the total $R_s$ of the solar cells
decreases to very low values, resulting in a high $FF$ (see Figure 3.15b). The strongly increasing $FF$ with high absolute values ($FF > 80\%$) proves that the power loss induced by recombination in the SCR ($j_{02}$ in the two-diode model) of the Ag front side metallization has only a minor influence on $FF$. The impact of recombination at the Ag metal contacts ($j_{01}$ in the two-diode model) is illustrated in Figure 3.15c, mainly decreasing $V_{oc}$. However, the relative loss in $V_{oc}$ is very low compared to the relative loss in $j_{sc}$ and $FF$ for the investigated range of $w_F$ and $d_F$. This also demonstrates that the recombination losses due to the Ag metallization are lower than the ones induced by resistance and shading.

![Graphs showing IV-characteristics](image)

Figure 3.15: IV-characteristics $j_{sc}$, $FF$, $V_{oc}$ and $\eta$ of RJ n-PERT solar cells with phosphorus s-FSF, extracted from the computed IV-curves (two-diode model), as a function of finger width $w_F$ and finger spacing $d_F$ of the Ag front side metallization H-grid. Note the non-linear scaling of $FF$ and $\eta$.

Although the influence of recombination at the metal contacts is low, it leads to a noticeable shift of the ideal $w_F$ and $d_F$ towards lower values, compared to the study without considering recombination losses (presented in Figure 3.14). Figure 3.15d illustrates that the highest $\eta$ can be achieved for thin fingers with $w_F \approx 40 \mu m$ and $d_F = 1.4 - 1.9 \ mm$. However, narrowing of the metal fingers, e.g. due to technical issues during screen-printing, results in a strong increase
of $R_S$ and therewith a strong decrease of $\eta$ in this region of $w_F$. It turns out that a much more insusceptible region, with comparable values in $\eta$, can be achieved by screen-printing Ag metal fingers with $w_F \approx 60 \, \mu\text{m}$ and $d_F = 2.0 – 2.2 \, \text{mm}$, leading to a higher process stability. These high values of $d_F$ are a consequence of the lateral carrier transport between the metal fingers which also occurs in the n-type base.

Based on this computation, a screen with mesh parameters 400/18 using tapered fingers (screen-opening of 45 - 65 $\mu\text{m}$) was used for front side metallization of the bifacial and mono-facial n-PERT solar cells investigated in this work.

### 3.4 Rear side metallization

In the previous section, it was demonstrated that screen-printing of Ag metal pastes for contact formation on the phosphorus-doped s-FSF of n-PERT solar cells results in metal contacts with high electrical quality, almost independent of the contact firing conditions. In this section, the much more challenging issue of contact formation on boron-doped emitters is discussed. Contact formation on boron emitters by means of screen-printing of metal pastes has been intensely investigated in recent years by several research groups [61, 134, 135, 138, 139, 157, 160-163]. It was shown in [134, 135], that the use of pure Ag pastes can lead to a high specific contact resistance $\rho_c$, resulting in substantial power loss due to an increased series resistance $R_S$ (see section 2.3.3). This is in agreement with the experimental results obtained in this work, using a pure Ag metal paste of the older generation for contacting boron emitters, as presented in the next section. Therefore, new pastes have been developed by adding a small amount of Al into the Ag pastes. By this means, $\rho_c$ can be significantly reduced, as first reported in [134] and later in [135, 138].

However, the use of Ag/Al pastes suffers from several disadvantages such as the formation of metal spikes which can potentially penetrate the SCR, as reported in [61, 135, 138-140], resulting in essential power loss due to recombination at the metal contacts and in the SCR. This power loss can potentially compensate the benefits of n-type Si substrates. Another drawback is the slightly higher line resistivity $r_l$ of Ag/Al pastes, compared to pure Ag pastes, because of the lower conductivity of Al.

In this section, the influence of contact formation on boron emitters by screen-printing of Ag/Al metal pastes on the performance of bifacial n-PERT Si solar cells is reviewed. Although intense research has been conducted in recent years, the contact formation mechanisms are not yet completely understood. In order to enhance the energy conversion efficiency $\eta$ of the solar cells, these mechanisms have to be understood and the limiting recombination at the Ag/Al metal contacts has to be minimized. Therefore, a detailed study of the emitter shunting behavior due to the formation of Ag/Al spikes during contact firing and its influence on the IV-characteristics of the bifacial n-PERT Si solar cells are presented in the following. These investigations are assisted by a detailed microstructural SEM analysis.
It has to be mentioned that current research indicates that it is possible to form appropriate metal contacts on boron-doped emitters by utilizing an optimized glass composition of pure Ag pastes [136, 137].

3.4.1 Contact resistance of rear side metallization

In a first experiment, $\rho_c$ of screen-printed Ag/Al metal contacts on boron emitters, formed by means of co-diffusion from PECVD doping layers, was measured on test-structures with a TLM pattern according to Figure 3.6. As reference, $\rho_c$ of screen-printed Ag and Ag/Al contacts was also measured on BBr$_3$-based boron emitters using the same TLM test-structure. The process sequence was kept similar to the bifacial solar cell process (described in section 3.1) in order to ensure the same conditions during contact firing.

Figure 3.16 illustrates the cross-section of the TLM test-structure. As base material, n-type Cz-Si substrates with $\rho_b \approx 6 \ \Omega\cdot\text{cm}$ were used. The wafers obtained an alkaline texture on both sides. After chemical cleaning, the boron emitter and the phosphorus-FSF were formed in one high temperature co-diffusion step in a POCl$_3$ tube furnace using a two-step process, as described in section 2.5.1, respectively by BBr$_3$ diffusion for the reference wafers. Subsequently, the boron emitter was passivated by a stack of Al$_2$O$_3$ (10 nm) and SiN$_x$:H (65 nm). The emitter sheet resistivities were varied in the range of $R_{sh} = 55 – 80 \ \Omega/\text{sq}$ for the co-diffused PECVD boron emitters, whereas $R_{sh} \approx 50 \ \Omega/\text{sq}$ was used for the BBr$_3$-based boron-emitter. Contact formation was conducted in an IR conveyor belt furnace for a wide range of set peak firing temperatures $T_S$ of 800 - 920°C.

![Figure 3.16: Schematic cross section of the TLM test structure used for the measurement of the specific contact resistance of screen-printed Ag and Ag/Al metal contacts on boron emitters.](image)

Optical properties like surface morphology and ARC, emitter/FSF doping profiles, thickness of the wafers and the amount of metal paste all influence the peak wafer temperature $T_W$ during contact firing [164]. $T_W$ is significantly lower than $T_S$. To understand the contact formation mechanisms, $T_W$ has to be known instead of $T_S$. A thermocouple was used to determine $T_W$ during contact firing for a given $T_S$ in a range of 590 – 960°C. The values of $T_W$, extracted from the measured temperature profiles (see examples of the $T_W$ profile in Figure 3.17), for a belt speed of 6 m/min are shown in Table 4. The temperature difference $\Delta T$ between $T_S$ and $T_W$ is around 40 – 55°C. The estimated measurement error is less than 10°C.
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Figure 3.17: Measured wafer temperature $T_w$ during contact firing in an IR conveyor belt furnace shown exemplary for $T_S = 920^\circ$C and $T_S = 800^\circ$C at a belt speed of 6 m/min.

Table 4: Measured peak wafer temperature $T_w$ for a set peak firing temperature $T_S$ at a belt speed of 6 m/min. The difference $\Delta T$ between $T_S$ and $T_w$ is also shown. The values in round brackets indicate the measurement error, if multiple temperature profiles were measured for a given $T_S$.

<table>
<thead>
<tr>
<th>$T_S$ [°C]</th>
<th>590</th>
<th>670</th>
<th>690</th>
<th>720</th>
<th>760</th>
<th>800</th>
<th>840</th>
<th>880</th>
<th>920</th>
<th>960</th>
</tr>
</thead>
<tbody>
<tr>
<td>$T_w$ [°C]</td>
<td>550(6)</td>
<td>616</td>
<td>645(7)</td>
<td>675</td>
<td>705</td>
<td>751(3)</td>
<td>793(8)</td>
<td>832(4)</td>
<td>867</td>
<td>919</td>
</tr>
<tr>
<td>$\Delta T$ [°C]</td>
<td>40</td>
<td>54</td>
<td>45</td>
<td>45</td>
<td>55</td>
<td>49</td>
<td>47</td>
<td>48</td>
<td>53</td>
<td>41</td>
</tr>
</tbody>
</table>

Figure 3.18a shows the specific contact resistance $\rho_c$ for the BBr$_3$ reference wafers. The use of the commercial pure Ag paste P3 leads to drastically high $\rho_c$ values above 100 mΩcm$^2$. This would lead to a detrimentally high relative power loss, attributed to the contact resistance, of $P_c > 18\%$ (equation (3.27)). The application of commercial Ag/Al paste P4 enables $\rho_c$ values below 5 mΩcm$^2$. Figure 3.18b shows the $\rho_c$ values measured on the co-diffused PECVD boron-doped emitters as function of $T_S$ for the different sheet resistivities. In general, $\rho_c$ decreases with increasing $T_S$ except for 920°C, where the determination of the contact resistance might be affected by strong shunting of the emitter and/or the formation of a thick non-conductive glass layer between the physical metal contact and wafer surface, which acts as tunneling barrier for the charge carriers [160]. Moreover, $\rho_c$ increases strongly for increasing $R_{sh}$. This can be explained by a decreased surface doping concentration $N_S$ and/or a shallower emitter profile with low emitter depth $d_e$ for high $R_{sh}$ values. It was demonstrated in [165] that $\rho_c$ is inversely proportional to $N_S$ and $d_e$, which means that lowering $N_S$ and/or $d_e$ results in increased $\rho_c$ values. The measurements reveal $\rho_c$ values of 3 - 9 mΩcm$^2$ for the screen-printed Ag/Al contacts on 55 – 70 Ω/sq PECVD boron emitters for low $T_S$. This leads to an increase in the relative power loss, attributed to the contact resistance, of 0.5 – 1.5%$_{abs}$ according to equation (3.27), compared to the values obtained for Ag pastes presented in section 3.3.3. In this case, the contact resistance causes a measurable contribution to the total power loss induced by the metallization
of around 7% (compare section 3.3.5). The measurement on the 70 Ω/sq (cell) emitter was conducted on a finished solar cell instead of TLM structure with slightly underestimated values of $\rho_c$. Since no difference was observed in $\rho_c$ values for the BBr$_3$-based and co-diffused PECVD boron emitters with similar doping profile ($R_{sh} \approx 50$ Ω/sq), it is assumed that the contact formation is independent of the boron containing doping source.

**Figure 3.18:** Specific contact resistance $\rho_c$ for screen-printed contacts: a) with commercially available Ag and Ag/Al pastes on BBr$_3$-based boron emitters as reference and b) with commercially available Ag/Al paste on co-diffused PECVD boron emitters. The measurement on the 70 Ω/sq (cell) emitter was conducted on a finished solar cell instead of TLM structure. The lines are guides for the eye.

In a second experiment, the electrical quality of Ag/Al pastes of the newer generation were tested in terms of contact resistance. Additionally, the temperature range was lowered, since it was evident during the studies on the formation of Ag/Al crystallites that $T_S < 820^\circ$C can be beneficial for the performance of bifacial n-PERT Si solar cells. This is further explained in section 3.4.3. TLM test-structures were produced according to the previous experiment. The contact formation of screen-printed Ag/Al contacts on a heavily doped ($R_{sh} \approx 50$ Ω/sq) as well as on a lowly doped ($R_{sh} \approx 135$ Ω/sq) BBr$_3$-based boron emitter were investigated. In addition, the boron emitter from an optimized co-diffusion with $R_{sh} \approx 80$ Ω/sq was tested. Both pastes show low $\rho_c$ values over the whole temperature range on the 50 Ω/sq BBr$_3$ emitter. Paste P6 exhibits significantly lower values for $\rho_c$ for the lowly doped emitters. Especially the 80 Ω/sq PECVD boron emitter can be contacted very well for $T_S \geq 800^\circ$C with paste P6, contrary to paste P4 of the older generation (compare Figure 3.19b). Based on these results, paste P6 was chosen for the fabrication of bifacial n-PERT Si solar cells, enabling $\rho_c < 5$ m$\Omega$cm$^2$ for $T_S$ of around 800°C.

However, a low $\rho_c$ is not the only requirement for an appropriate screen-printed contact. In particular, when using Al containing pastes shunting problems due to Ag/Al spike formation may occur if the firing process is not controlled carefully. This is discussed in section 3.4.3.
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**Figure 3.19:** Specific contact resistance $\rho_c$ for screen-printed contacts on BBr$_3$-based and co-diffused PECVD boron emitters: a) with commercial Ag/Al paste P5 and b) with commercial Ag/Al paste P6. The lines are guides for the eye.

### 3.4.2 Line resistivity of rear side metallization

Another relevant parameter for describing the electrical quality of screen-printed Ag/Al metal contacts is the line resistance $R_l$, which also contributes to $R_S$ (compare section 2.3.3). As mentioned above, $R_l$ depends on the width and height of the screen-printed metal fingers. As it is often not trivial to measure these values, the line resistivity $r_l$ can be determined alternatively. $r_l$ includes all information about the cross section of the Ag/Al metal fingers. The shape of the fingers, and therefore $r_l$, depends strongly on the parameters of the mesh, the composition of the metal paste and the screen-printing parameters, as demonstrated in section 3.3.1 for screen-printed Ag metal contacts. A variation of the mesh parameters was not performed in this study, since Ag/Al pastes are based on pure Ag pastes. As the screen-printing of Ag/Al metal pastes was conducted for rear side metallization, it is not crucial to use narrow fingers. For rear side metallization of mono-facial Si solar cells, shading is a negligible issue since the illumination is solely conducted from the front side. Although this is also almost true for bifacial solar cells, the power gain of the bifaciality decreases for increasing metallization fraction on the rear side. The exact composition of the commercial Ag/Al pastes used within this work are not known because it is a corporate secret. However, it is believed that the paste comprises pure Ag with the addition of around 5% Al [165]. It turned out in this work that the Al particles tend to clog the mesh of the screen during screen-printing, resulting in narrowing and even interruptions of the Ag/Al metal fingers. Therefore, a mesh with the parameters 300/20 was used instead of a mesh with the parameters 400/18 (used for the narrow front side Ag fingers) to lower the amount of wires per inch, resulting in wider mesh openings and a lower probability of getting clogged during screen-printing. The spacing between substrate and screen (1.8 mm), squeegee pressure (1.2 bar) and printing velocity (50 mm/s) were kept constant during the whole study.

The measurement of the line resistivity $r_l$ was performed on the same TLM structures used for the determination of $\rho_c$. The values of $r_l$ as a function of the screen opening are depicted in
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Figure 3.20. The experimentally observed broadening of the Ag/Al fingers is almost the same as the one observed for the Ag fingers for front side metallization (reviewed in section 3.3.2). In Figure 3.20a, the difference in \( r_l \) depending on the type of boron emitter was investigated. As expected, no difference can be observed between BB\( \text{Br}_3 \) and PECVD boron emitters for paste P4. In Figure 3.20b, the difference in \( r_l \) for Ag/Al pastes of the newer generation was studied. Although paste P6 shows the best contact behavior in terms of a low \( \rho_c \), it exhibits slightly higher \( r_l \) values compared to pastes P4/P5. This can be explained by an optimized paste composition for ideal contact formation, suffering from slightly increased line resistivity.

![Figure 3.20](image.png)

**Figure 3.20:** Line resistivity \( r_l \) in dependence of the screen opening for different commercial Ag/Al pastes. In a) a comparison of \( r_l \) for paste P4 is given on a BB\( \text{Br}_3 \) and PECVD boron emitter. In b) the Ag/Al pastes of the newer generation are shown for smaller screen openings. The lines are guides for the eye.

![Figure 3.21](image.png)

**Figure 3.21:** Influence of the set peak firing temperature \( T_s \) on the line resistivity \( r_l \). The lines are guides for the eye.

The next investigation was focused on a possible dependence of the line resistivity \( r_l \) on the set peak firing temperature \( T_s \). Therefore, TLM test-structures were prepared and fired at \( T_s \) in a
range of 820 - 920°C at a belt speed of 6 m/min. The results are shown in Figure 3.21. In the scope of the measurement uncertainties (error bars in Figure 3.21), no noticeable impact of $T_3$ on $r_f$ can be observed. This observation corresponds to the observation on pure Ag pastes.

Although, screen-printed Ag/Al metal contacts show an appropriate electrical quality, in a next step the influence of the Ag/Al metal contacts on the IV-characteristics of the bifacial n-PERT solar cells, potentially caused by power losses due to recombination at the metal contacts, must be analyzed and quantified.

### 3.4.3 Ag/Al metal spike formation

In the following, a detailed microstructural investigation of screen-printed Ag/Al metal contacts by means of SEM analysis is presented. In this work, the Neon 40 EsB thermal field emission SEM from the manufacturer Zeiss was used for top-view images of the contact spots. The SEM is additionally equipped with a focused ion beam (FIB) system, which allows the modification of the wafer surface. The FIB was used for the preparation of cross-sections of the metal contact spots, enabling measurements of the contact depth. The study on the formation of Ag/Al metal spikes was conducted on bifacial n-PERT Si solar cells, corresponding to the process sequence described in section 3.1, except for the EB step.

As base material, n-type Cz-Si (6 Ωcm) was utilized. The wafers were alkaline textured on both sides. Boron emitter and phosphorus-doped FSF were formed in one single high temperature co-diffusion step in a POCl₃ tube diffusion furnace, using a two-step process as described in section 2.5.1. Four different co-diffused boron emitters with $R_{sh} = 40 - 70$ Ω/sq (measured by a 4PP setup) were formed to study the impact of the Ag/Al spikes on the IV-characteristics of the solar cells in dependence of the emitter profile. The boron doping profiles, determined by ECV measurements (outlined in section 2.5.4), are illustrated in Figure 3.22. The ECV measurements revealed emitter depths $d_e$ in the range of 500 – 700 nm and surface doping concentrations $N_S$ of 3 - 4·10¹⁹ cm⁻³. For all solar cells, a highly-doped homogeneous phosphorus FSF with $R_{sh} = 60$ Ω/sq was formed (without EB) to keep the process simple and to guarantee screen-printed Ag front side metal contacts with high electrical quality in terms of low $\rho_c$, which is ideally independent of the set peak firing temperature $T_3$. To investigate a potential influence of the dielectric passivation layers on the Ag/Al spike formation, two different passivation layer stacks, consisting of PA-ALD Al₂O₃ (10 nm) and SiNₓ:H (65 nm) or thermal SiO₂ (7 nm) and SiNₓ:H (68 nm), were used for passivation of the boron emitters. The phosphorus FSF was passivated by a stack of thermal SiO₂ (7 nm) and SiNₓ:H (68 nm) in all cases.

Contact formation was performed by screen-printing the commercial Ag paste P1 on the FSF and the commercial Ag/Al paste P6 on the boron emitter. Contact firing was conducted in an IR conveyor belt furnace with $T_3$ in the extended range of 720 – 960°C. The characteristic IV-curves of the bifacial n-PERT solar cells were measured using a flasher from the manufacturer h.a.l.m, which is equipped with an absorbing black back sheet. Non-metalized samples were
simultaneously processed similar to the solar cells, in order to examine the influence of $T_S$ on passivation quality. The solar cell results are presented in section 3.4.4.

![Doping profiles of the boron emitters manufactured by means of co-diffusion from a PECVD SiO$_2$:B doping layer, measured by an ECV setup. Published in [91].](image)

**Figure 3.22:** Doping profiles of the boron emitters manufactured by means of co-diffusion from a PECVD SiO$_2$:B doping layer, measured by an ECV setup. Published in [91].

Subsequently, one part of the solar cells was prepared for the microstructural SEM analysis. The cells were cut into small samples. The samples for the top-view SEM images were etched in an aqueous solution of HF (5%) to remove the glass layer as well as the dielectric layer stack between the metal contact and the sample surface. Thereby, the bulk of the Ag/Al metal contacts separates from the wafer surface. Only the metal spikes, which were covered by the glass layer, remain on the wafer surface. In order to additionally remove the metal spikes, the samples are etched in *aqua regia*, consisting of HCl and HNO$_3$. After this etching sequence, the imprints of the metal spikes on the wafer surface are visible.

Figure 3.23 illustrates the top-view SEM images after removal of the Ag/Al paste and metal spikes with an etching sequence of HF and *aqua regia* for samples fired at different $T_S$. It is clearly visible, that the amount and size of the Ag/Al spikes (visible as inverted pyramids marked in light blue), increase substantially for increasing $T_S$. This observation coincides with the decreasing specific contact resistance $\rho_c$ for increasing $T_S$, as discussed in section 3.4.1. This can be explained by the larger contact area for the increasing amount and size of the Ag/Al spikes for increasing $T_S$. At $T_S \geq 890^\circ$C, massive Ag/Al spikes appear. Unfortunately, it is not possible to determine reliable values for the depth of the metal spikes from the top-view SEM images. Therefore, cross sections of the Ag/Al contact spots were prepared by FIB.

The samples for the cross-section images were etched back in an aqueous solution of HF (5%) to remove the glass layer as well as the dielectric layer stack between the metal contact and the sample surface. The Si surface with the Ag/Al contact spots is revealed. To determine the depth of the Ag/Al spikes, cross-sections of the contact spots were prepared by FIB. Figure 3.24
shows the SEM images of the cross-sections of the Ag/Al spikes for $T_S = 800 - 960^\circ$C. The depth of the Ag/Al spikes is marked with light blue lines.

![SEM images of the cross-sections of the Ag/Al spikes](image)

**Figure 3.23**: Top-view SEM images of samples etched with a sequence of an aqueous solution of HF (5%), in order to remove the glass layer as well as the dielectric layer stack between the Ag/Al metal contact and the sample surface, and *aqua regia* in order to additionally remove the metal spikes. The Ag/Al spikes are visible as inverted pyramids (marked in light blue). The amount and size of the Ag/Al spikes increases strongly for increasing set peak firing temperature $T_S$. Published in [92].

In order to give a statistical conclusion on the depth of the Ag/Al metal spikes, a large number cross-sections of contact spots were prepared and analyzed for each $T_S$, similar to the procedure described in [93]. Figure 3.25 depicts a boxplot of the Ag/Al spike depths for each $T_S$. The mean Ag/Al spike depths, as well as the deviation of the values increase strongly for increasing $T_S$. The SEM analysis reveals spike depths $> 1 \mu$m in average, even for low $T_S$ of 800°C. This confirms, that the Ag/Al spikes are already deep enough to penetrate through the SCR. Further clarification is evident by observing the doping profiles of the boron emitters illustrated in Figure 3.22, which exhibit emitter depths in the range of 500 – 700 nm. In general, lowering $T_S$ causes a decreased spike area density and therefore less metal spikes penetrating through the SCR. On the other hand, lowering the area density of the metal spikes results in an increase in $\rho_c$. Therefore, a compromise in the firing conditions between appropriate contact formation in terms of low $\rho_c$ and density of Ag/Al spikes has to be made. To achieve optimal contact firing
conditions, a deeper understanding of the impact of the Ag/Al metal spikes on the solar cell characteristics is mandatory.

Figure 3.24: SEM images of FIB cross-sections of the Ag/Al contact spots for different set peak firing temperatures $T_S$. The samples were etched in an aqueous solution of HF (5%) to remove the glass layer as well as the dielectric layer stack between the metal contact and the sample surface. The depths of the Ag/Al spikes are marked with light blue lines. Area density and average depth of the spikes increase strongly with increasing $T_S$. Published in [91].

Figure 3.25: Depths of the metal spikes determined by SEM images of FIB cross sections of the contact spots. The SEM analysis revealed spike depths > 1 μm in average, even for low $T_S$ of 800°C. This confirms, that the Ag/Al spikes are already deep enough to penetrate the SCR. Published in [91].
3.4.4 Influence of Ag/Al metal spikes on solar cell characteristics

The IV-characteristics of the bifacial n-PERT solar cells with various co-diffused boron emitter profiles (shown in Figure 3.22) and passivation layer stacks, measured with a flasher device using an absorbing black rear side, are depicted in Figure 3.26. Note that the presented values are averaged over three solar cells and that the error bars correspond to the standard deviation. The solar cell energy conversion efficiency $\eta$ is continuously decreasing for increasing set peak firing temperature $T_S$ (for $T_S \geq 760^\circ$C) for all boron emitter profiles and passivation layer stacks, as depicted in Figure 3.26a. Note that the measured peak temperature of the solar cells $T_W$ during contact firing is around 40 – 55°C lower than $T_S$ (see Table 4 in section 3.4.1). It is observed that the decrease in $\eta$ is more distinct for the solar cells with Al$_2$O$_3$/SiN$_x$ passivation layer stack (green colored triangles in Figure 3.26a), compared to the solar cells with SiO$_2$/SiN$_x$ passivation layer stack for boron emitter passivation. In the following, the origin of this $\eta$ loss increasing with $T_S$ is analyzed and the ideal Ag/Al metallization H-grid is computed, on the basis of the presented experimental data, in order to minimize the mettallization-induced power loss.

The short circuit current density $j_{sc}$ is nearly constant for increasing $T_S$ (only slightly decreasing for the solar cells with Al$_2$O$_3$/SiN$_x$ passivation layer stack), except for $T_S \leq 720^\circ$C and $T_S \geq 960^\circ$C (see Figure 3.26b). This behavior can be explained by insufficient contact formation for $T_S \leq 720^\circ$C, resulting in detrimentally high values of the specific contact resistance $\rho_c$. This agrees with the examinations on contact formation of screen-printed Ag and Ag/Al metal pastes for front and rear side metallization, presented in the sections 3.3.3/3.4.1, and can mainly be attributed to the screen-printed Ag/Al contacts. This is based on the fact that a highly-doped 60 $\Omega$/sq phosphorus FSF was used, which allows for low $\rho_c$, being almost independent of $T_S$. The strong decrease in $j_{sc}$ for $T_S \geq 960^\circ$C can be explained by melting and agglomeration of both metal pastes, which leads to a high line resistivity $r_l$. These statements are confirmed by the drastic increase of the series resistance $R_S$ in these temperature zones ($T_S \leq 720^\circ$C and $T_S \geq 960^\circ$C), as depicted in Figure 3.26e. Note that $j_{sc}$ is not affected by $R_S$ until $R_S$ exhibits values of around 10 $\Omega$cm$^2$ or higher, as discussed in section 2.3.4.

Figure 3.26c reveals that $V_{oc}$ is continuously decaying for increasing $T_S$ for all boron emitter profiles and passivation layer stacks. Since the parallel (shunt) resistance $R_P$ is above 1 k$\Omega$cm$^2$ for all $T_S$ (see Figure 3.26f), its influence on $V_{oc}$ is of minor relevance. Thus, the continuously decreasing $V_{oc}$ for increasing $T_S$ can only be caused by an increasing saturation current density of the first and/or second diode $j_{01}$/$j_{02}$, according to the two-diode model. The influence of $j_{01}$ and $j_{02}$ on $V_{oc}$ has been outlined in section 2.3.4. The $j_{01}$ increase must be attributed to the screen-printed Ag/Al metal contacts on the boron emitter but also to the screen-printed Ag metal contacts on the phosphorus-doped FSF (the $j_{01}$ values of Ag metal contacts have been reviewed in section 3.3.4), whereas the increase of $j_{02}$ is expected to be caused primarily by the Ag/Al metallization on the boron emitter. Furthermore, the decay of $V_{oc}$ may also be explained partly by a degradation of the passivation quality for increasing $T_S$. It has been shown in [124] that the passivation quality of Al$_2$O$_3$ layers is strongly reduced by high temperature treatments, which
results in an increased $j_{01}$ of the passivated area. To study the influence of contact firing on the passivation quality of the Al$_2$O$_3$/SiNx and SiO$_2$/SiNx passivation layer stacks, non-metallized samples were processed similar to the solar cells and fired at the same $T_S$ used for contact firing of the solar cells.

**Figure 3.26:** IV-characteristics of the bifacial n-PERT Si solar cells measured with a flasher tool as a function of the set peak firing temperature $T_S$ for various co-diffused PECVD boron emitter profiles and passivation layer stacks. Note that the values are averaged over three solar cells each and the error bars correspond to the standard deviation. The solid lines serve as guides to the eye. Published in [91].
Figure 3.27a shows the evolution of $V_{oc}$ and implied $V_{oc}$ ($iV_{oc}$) for the 64 $\Omega$/sq (Al$_2$O$_3$/SiNx passivation layer stack) and the 70 $\Omega$/sq (SiO$_2$/SiNx passivation layer stack) co-diffused boron emitters. The $iV_{oc}$ values were quantified by the quasi steady state photoconductance decay method using a Sinton WCT-120 lifetime tester. The passivation quality of the Al$_2$O$_3$/SiNx:H passivation stack degrades considerably for increasing $T_S$, whereas the SiO$_2$/SiNx:H passivation layer stack shows up stable over the whole range of $T_S$ (compare $iV_{oc}$ curves, filled green triangle and filled blue diamond symbols in Figure 3.27a). Although the passivation quality of the SiO$_2$/SiNx:H passivation stack was very poor in this experiment, probably because of process issues, these observations agree well with investigations of other research groups, e.g. reported in [111]. $V_{oc}$ of the solar cells decays stronger for increasing $T_S$ when applying an Al$_2$O$_3$/SiNx:H passivation stack, compared to the use of an SiO$_2$/SiNx:H passivation stack for boron emitter passivation, due to the mentioned degradation effect during contact firing. This also explains the slight decrease of $j_{sc}$ of the solar cells with Al$_2$O$_3$/SiNx:H passivation layer stack, whereas $j_{sc}$ of the solar cells with SiO$_2$/SiNx:H passivation layer stack for passivation of the boron emitter remains almost constant over the whole $T_S$ range (see Figure 3.26b).

**Figure 3.27:** a) Evolution of $V_{oc}$ and implied $V_{oc}$ for the 64 $\Omega$/sq (Al$_2$O$_3$/SiNx, passivation layer stack) and the 70 $\Omega$/sq (SiO$_2$/SiNx, passivation layer stack) co-diffused boron emitters. b) Difference $iV_{oc} - V_{oc}$, which equals the metallization-induced voltage losses of the solar cells at open-circuit conditions. The metallization-induced losses are strongly increasing for increasing set peak firing temperature $T_S$ and independent of the passivation layer stack. Note that the values are averaged over three solar cells (or non-metallized samples) each and the error bars correspond to the standard deviation. The solid lines serve as guides to the eye. Published in [91].

The difference between $iV_{oc}$ and $V_{oc}$, which equals the metallization-induced voltage loss of the solar cells at open-circuit conditions, is shown in Figure 3.27b. This consideration is based on the fact that $iV_{oc}$ corresponds to the $V_{oc}$ a solar cell would have without metallization and the thereby induced power loss due to recombination. The difference in the wafer peak temperature $T_w$ between the non-metallized $iV_{oc}$ samples and solar cells during the contact firing step is assumed to be small, because of the low metallization fraction of the solar cells of less than 7%
per side. It is visible that the total metallization-induced loss of $V_{oc}$ is in the range of 20 - 90 mV for both passivation layer stacks and follows the same trend (compare filled green square and filled blue circle symbols in Figure 3.27b). Based on these findings, the formation of Ag/Al metal spikes during contact firing and their impact on $V_{oc}$ of the solar cells for increasing $T_S$ seems to be independent of using either an Al$_2$O$_3$/SiN$_x$ or SiO$_2$/SiN$_x$ dielectric passivation layer stack. On the other hand, the IV-characteristics of the solar cells shown in Figure 3.26 indicate that the detrimental influence of Ag/Al metal spikes on the solar cell IV-characteristics is strongly dependent on the boron emitter profile of the solar cells.

In order to quantify and analyze the detrimental influence of Ag/Al metal spikes on the IV-characteristics of bifacial n-PERT solar cells in dependence of the boron emitter profile, the two-diode model (equation (2.8)) was fitted to the measured IV-curves of the solar cells shown in Figure 3.26. From these fits, the total values of $j_{01}$ and $j_{02}$ of the solar cells were extracted. Note that it was not possible to reliably fit the two-diode model to the measured IV-curves of the solar cells fired at $T_S = 720^\circ$C and $T_S = 960^\circ$C, due to the ohmic behavior of the IV-curves because of rigorous $R_S$ values. The total $j_{01}$ of the solar cells can also be calculated from $V_{oc}$ and $j_{sc}$ using the one-diode model as proposed by Fellmeth et al. in [166] by

$$j_{01} = \frac{j_{sc}}{e^{(\frac{qV_{oc}}{k_BT})} - 1}$$

(3.31)

if the influence of $j_{02}$ on $V_{oc}$ can be neglected (ideality factor $n$ of the one-diode model close to one). This calculation was used to cross check the $j_{01}$ values extracted from the two-diode model. Figure 3.28 illustrates the $j_{01}$ values extracted from the two-diode model and the ones calculated from $V_{oc}$ and $j_{sc}$ according to equation (3.31).

Although equation (3.31) gives roughly comparable values of $j_{01}$ for low $T_S$, it turned out that the influence of $j_{02}$ on $V_{oc}$ cannot be neglected at least for $T_S \geq 920^\circ$C (compare filled and blank symbols in Figure 3.28). In this high temperature region, the $j_{01}$ values calculated according to equation (3.31) deviate substantially from the $j_{01}$ values extracted from the two-diode model. Furthermore, it is observed that the difference between these values increases for decreasing $R_{th}$ of the boron emitter profiles. These observations are explained later in this section when reviewing the $j_{02}$ values of the solar cells.

To get a deeper understanding of $j_{01}$ as a function of $T_S$ and to calculate the contribution of the metallization to the total $j_{01}$ of the solar cells, $j_{0,2d}$ ($j_{01}$ extracted from the two-diode model), $j_{0,1Voc}$ ($j_{01}$ measured on non-metallized samples), linear fit-functions to model the $j_{01}$ curves and the difference between $j_{0,2d}$ and $j_{0,1Voc}$ are shown in Figure 3.29. Unfortunately, the $j_{0,1Voc}$ values were only measured for the 64 $\Omega$/sq (Al$_2$O$_3$/SiN$_x$ passivation layer stack) and the 70 $\Omega$/sq (SiO$_2$/SiN$_x$ passivation layer stack) boron emitters (blank symbols in Figure 3.29b/c). The linear fits seem reasonable since they lie within the error bars over the whole range of $T_S$. The slope of the total $j_{01}$ curve as a function of $T_S$ is highest for the solar cells with Al$_2$O$_3$/SiN$_x$ passivation layer stack (see Figure 3.29b). This value comprises the increase of the total $j_{01}$
caused by the metallization but also caused by the degradation of the Al₂O₃/SiNx passivation layer stack for increasing $T_s$.

**Figure 3.28:** Saturation current density of the first diode $j_{01}$ extracted from the two-diode model and calculated from $V_{oc}$ and $j_{sc}$ (according to equation (3.31)) for a) 40/50 $\Omega$/sq (SiO₂/SiNx passivation layer stack) co-diffused boron emitters and b) 64 $\Omega$/sq (Al₂O₃/SiNx passivation layer stack) and 70 $\Omega$/sq (SiO₂/SiNx passivation layer stack) co-diffused boron emitters. Note that the values are averaged over three solar cells each and the error bars correspond to the standard deviation. The solid lines serve as guides to the eye.

In order to determine the impact of the metallization only, the difference between $j_{01,Voc}$ and $j_{01,2d}$ was calculated for the 64 $\Omega$/sq (Al₂O₃/SiNx passivation stack) and the 70 $\Omega$/sq (SiO₂/SiNx passivation stack) boron emitters (see Figure 3.29d). This consideration is based on the fact that $j_{01,Voc}$ corresponds to the total $j_{01}$ value a solar cell would have without metallization. Since the SiO₂/SiNx passivation stack hardly suffers from degradation during contact firing, the slope of the $j_{01,Voc} - j_{01,2d}$ curve approximates the slope of the $j_{01,2d}$ curve for the solar cells with 70 $\Omega$/sq boron emitter (compare fit functions for the filled blue diamonds in Figure 3.29c/d). Therefore, it is assumed that the slope of the $j_{01,2d}$ curves of the solar cells with the 40/50 $\Omega$/sq boron emitters would also approximate the slope of the $j_{01,Voc} - j_{01,2d}$ curves, since the same SiO₂/SiNx passivation layer stack was used.

When comparing only the solar cells with SiO₂/SiNx passivation stack, it turns out that the slope of the $j_{01}$ curve is lower for the boron emitter profiles with lower $R_{sh}$ (compare Figure 3.29a/c). This indicates that the high doping of the boron emitter profiles with low $R_{sh}$ (deep emitters with high surface doping concentration) effectively shields the recombinative metal contacts and thereby hinders the minority charge carriers from recombining. This electrical shielding effect also explains why the increasing area density of the Ag/Al metal spikes for increasing $T_s$ does not affect the $j_{01}$ values as much as observed for the lower doped boron emitter profiles. Although it is expected, on the basis of these findings, that the $V_{oc}$ of the solar cells with boron emitters with low $R_{sh}$ is less decaying for increasing $T_s$ than the solar cells with boron emitters.
with higher $R_{sh}$, because of the electrical shielding effect, it is observed that the $V_{oc}$ of the solar cells with the boron emitter with lowest $R_{sh}$ decays strongest (see black squares in Figure 3.26c). This gives a first indication that an increasing $j_{02}$ for increasing $T_S$ must also be considered for explaining the total $V_{oc}$ loss of the solar cells. Though, the major influence of an increased $j_{02}$ is a reduction of $FF$, it can also lead to a loss of $V_{oc}$, as discussed in section 2.3.4. Whether or not an increased $j_{02}$ can explain the observed trend of $V_{oc}$ as a function of $T_S$ for the different boron emitter profiles and its impact on $FF$ is studied in the following.

![Figure 3.29: a) Saturation current density of the first diode extracted from the two-diode model $j_{01,2d}$ for the 40/50 $\Omega$/sq (SiO$_2$/SiN$_x$) passivation) boron emitters, $j_{01,2d}$ and $j_{01,iVoc}$ ($j_{01}$ measured on non-metallized samples) are shown in b) for the 64 $\Omega$/sq (Al$_2$O$_3$/SiN$_x$) passivation and in c) for the 70 $\Omega$/sq (SiO$_2$/SiN$_x$) passivation) boron emitters. In d) the difference between $j_{01,iVoc}$ and $j_{01,2d}$ is depicted for the 64 $\Omega$/sq (Al$_2$O$_3$/SiN$_x$) passivation and the 70 $\Omega$/sq (SiO$_2$/SiN$_x$) passivation) boron emitters, which corresponds to the contribution of the metallization to the total $j_{01}$ of the solar cells. Note that the values are averaged over three solar cells (or non-metallized samples) each and the error bars correspond to the standard deviation. The solid lines serve as guides to the eye and the dashed lines are linear fit functions to model the data.

$FF$ is directly linked to $R_S$, as also outlined in section 2.3.4. The decrease of $R_S$ for increasing $T_S$ (see Figure 3.26e) is primarily related to the decrease of $\rho_c$ due to an increase of the area.
density of the Ag/Al metal spikes, as shown in the previous section. Although the decreasing $R_S$ should result in an increase in $FF$ for increasing $T_S$, a noticeable decrease in $FF$ is observed instead for all boron emitter profiles and passivation layer stacks (see Figure 3.26d). This might be caused by a considerable increase of $j_{02}$ for increasing $T_S$, potentially overcompensating the effect of the decreasing $R_S$ because of its compromising influence on $FF$ (see section 2.3.4). The enhanced $j_{02}$ might be induced by the increase of the area density and depth of the Ag/Al metal spikes for increasing $T_S$. The Ag/Al metal spikes are deep enough to easily penetrate the SCR of all of the presented boron emitter profiles, even for low $T_S$, as reviewed in the previous section. This leads to an increased recombination activity in the SCR, depending strongly on the area density of the Ag/Al metal spikes and thus on $T_S$. To clarify this issue, a fill factor loss analysis is presented in the following.

Figure 3.30 shows the $j_{02}$ values extracted from the fits of the two-diode model to the measured IV-curves of the bifacial n-PERT solar cells presented in Figure 3.26. The $j_{02}$ data is fitted by exponential fit functions to model the trend of $j_{02}$ for increasing $T_S$. It is evident that the absolute values of $j_{02}$ and its increase for increasing $T_S$ are higher for the boron emitter profiles with lower $R_{sh}$.

![Figure 3.30](image_url)

Figure 3.30: Saturation current density of the second diode $j_{02}$ extracted from the two-diode model a) for the 40/50 $\Omega$/sq (SiO$_2$/SiN$_x$ passivation layer stack) co-diffused boron emitters and b) for the 64 $\Omega$/sq (Al$_2$O$_3$/SiN$_x$ passivation layer stack) and 70 $\Omega$/sq (SiO$_2$/SiN$_x$ passivation layer stack) co-diffused boron emitters. The dashed lines correspond to the exponential fit functions used to model the data. Note that the values are averaged over three solar cells each and the error bars correspond to the standard deviation. Part of the data is published in [91].

A high $j_{02}$ results in a substantial loss of $FF$. The $FF$ loss attributed to $j_{02}$ can be quantified by the difference between the ideal $FF$ (calculated from $V_{oc}$ by equation (2.6)) and the pseudo $FF$ ($FF_0 – pFF$). $pFF$ was determined via Suns-Voc measurements, introduced in section 3.2.2. $FF_0 – pFF$ of the solar cells, as a function of $T_S$, is shown in Figure 3.31a. Note that $FF_0 – pFF$ equals the $FF$ loss induced by $j_{02}$ only if $R_P$ has no influence on $FF$. Since $R_P$ is above 1 k$\Omega$cm$^2$
for all emitter profiles over the whole range of $T_S$ (see Figure 3.26f), its influence on $FF$ is negligible. Therefore, the $FF$ loss shown in Figure 3.31a can primarily be attributed to enhanced recombination in the SCR for increasing $T_S$, causing an enhanced $j_{02}$. It is visible that this $j_{02}$-induced loss of $FF$ is continuously increasing for increasing $T_S$ for all boron emitter profiles.

According to the total values of $j_{02}$ shown in Figure 3.30, the total $j_{02}$-induced $FF$ loss and its increase for increasing $T_S$ is more distinct for boron emitter profiles with lower $R_{sh}$.

**Figure 3.31:** a) Difference between the ideal fill factor $FF_0$ (calculated by equation (2.6)) and the pseudo $FF$ (quantified by Suns-Voc measurements), which equals the loss of $FF$ induced by $j_{02}$ (if the influence of $R_p$ on $FF$ can be neglected). In b) the difference between $pFF$ and $FF$ is shown, which equals the loss of $FF$ induced by the series resistance $R_s$. In c) the total $FF$ loss is depicted and in d) the loss of $V_{oc}$ induced by recombination in the SCR is shown. Note that the values are averaged over three solar cells each and the error bars correspond to the standard deviation. The solid lines serve as guides to the eye.

As mentioned before, $R_s$ is a continuously decaying function of $T_S$ for all of the boron emitter profiles (see Figure 3.26e). Thus, the $FF$ loss induced by $R_s$ is expected to decay continuously for increasing $T_S$ also. Figure 3.31b illustrates the difference between $pFF$ and $FF$, which equals the loss of $FF$ caused by $R_s$. As expected, $pFF − FF$ is a monotonously decreasing function of $T_S$. This explains why the total loss of $FF$, which can be quantified by the difference between $FF_0$ and $FF$, is relatively flat for $T_S \leq 880^\circ C$ (see Figure 3.31c). Figure 3.31a/b reveal that the
FF loss induced by \( j_{o2} \) is in the same range \((2 - 4\%_{\text{abs}})\) as the one induced by \( R_s \) for \( T_s \leq 880^\circ\text{C} \) (except for the solar cells with the 40 \( \Omega/\text{sq} \) boron emitters, where the \( j_{o2} \)-induced FF loss is even higher). For \( T_s \geq 920^\circ\text{C} \), the FF loss is dominated by \( j_{o2} \) for all studied boron emitter profiles, overcompensating the effect of the decreasing \( p\text{FF} \) curve. These findings clarify that the FF loss due to recombination in the SCR, which is induced by the Ag/Al metal spikes penetrating the SCR, is the main contribution to the total FF loss of bifacial n-PERT solar cells for high \( T_s \). Even for low \( T_s \), the \( j_{o2} \)-induced FF loss is in the range of \( 2 - 4\%_{\text{abs}} \), strongly depending on \( R_{sh} \) of the boron emitter profile.

It was demonstrated earlier that \( j_{o1} \) is only slightly increasing for increasing \( T_s \) for the solar cells with boron emitter profiles with low \( R_{sh} \). Besides its strong detrimental impact on FF, it was stated that the increasing \( j_{o2} \) might also affect \( V_{oc} \) substantially. To prove this statement, the \( j_{o2} \)-induced \( V_{oc} \) loss was estimated from the two-diode fit by setting \( j_{o2} = 1 \text{nA/cm}^2 \), while keeping all other parameters fixed. This procedure allows to estimate the \( V_{oc} \) a solar cell would have for a negligible \( j_{o2} \). The difference between the measured \( V_{oc} \) and the \( V_{oc} \) extracted from the two-diode fit, assuming \( j_{o2} = 1 \text{nA/cm}^2 \) equals the \( j_{o2} \)-induced \( V_{oc} \) loss and is shown in Figure 3.31c. As expected, the \( j_{o2} \)-induced \( V_{oc} \) loss is higher for the solar cells with boron emitters with lower \( R_{sh} \). This agrees well with the total \( j_{o2} \) values (shown in Figure 3.30) and follows a similar trend as observed for the \( j_{o2} \)-induced FF loss (compare Figure 3.31a/d). Indeed, the influence of \( j_{o2} \) is smaller than the influence of \( j_{o1} \) on \( V_{oc} \) (compare total \( V_{oc} \) loss of 20 – 30 mV of the solar cells shown in Figure 3.27b with the \( j_{o2} \)-induced \( V_{oc} \) loss illustrated in Figure 3.31d for \( T_s \leq 880^\circ\text{C} \)). However, the \( j_{o2} \)-induced \( V_{oc} \) loss is nevertheless in the range of 5 – 10 mV for the solar cells with the 50/64/70 \( \Omega/\text{sq} \) boron emitters for \( T_s \leq 880^\circ\text{C} \), showing an increasing contribution to the total \( V_{oc} \) loss of the solar cells for decreasing \( R_{sh} \). If the solar cells are fired at \( T_s \geq 920^\circ\text{C} \) or if the 40 \( \Omega/\text{sq} \) boron emitter is applied, the \( j_{o2} \)-induced \( V_{oc} \) loss even exceeds 15 mV.

At first glance, it might be surprising that the above-presented detrimental impact of the Ag/Al metal spikes on FF and \( V_{oc} \) and therefore on \( \eta \) is higher for the solar cells with boron emitters with lower \( R_{sh} \), which exhibit deeper doping profiles with higher surface doping concentration and therefore have a better electrical shielding of the physical metal contacts. Because of the deeper doping profiles for the boron emitters with lower \( R_{sh} \), it seems plausible that the Ag/Al spikes have a lower chance of penetrating the SCR for these boron emitters. However, as found in section 3.4.3, the Ag/Al metal spikes are deep enough (> 1 \( \mu\text{m} \)) to easily penetrate the SCR of all of the studied boron emitter profiles, even for low \( T_s \) (see Figure 3.25). Since the Ag/Al spikes represent direct contacts between the highly-doped boron emitter and the metal, also known as Schottky-contacts [167], the current transport via these metal spikes depends on the potential barrier, also referred to as Schottky barrier. The current transport at Schottky-contacts, by thermionic emission over the Schottky barrier and thermionic field emission by tunneling through the Schottky barrier, depends on the temperature and the doping concentration of the boron emitter. The width of the Schottky barrier is reduced for a high boron emitter doping concentrations, as reported in [168]. This results in a higher probability for charge carriers to
tunnel through the potential barrier and contribute to parasitic leakage currents in the case of a metal spike penetrating the SCR for a highly-doped boron emitter. Since, as mentioned before, the Ag/Al metal spikes are deep enough to easily penetrate the SCR independent of $R_{sh}$ and emitter depth $d_e$ of all of the boron emitter profiles studied in this work, it is assumed that the detrimental leakage currents in the SCR are proportional to the boron doping concentration. This assumption is consistent with all of the findings presented in this section and agrees well with existing analytical models for the calculation of the specific contact resistance presented by Lohmüller et al. [163] and Fritz et al. [157], explaining the experimentally observed lower contact resistance for boron emitters with lower $R_{sh}$.

Note that the observed deviation of the detrimental influence of the Ag/Al metal spikes on the IV-characteristics of the bifacial n-PERT solar cells for the different boron emitter profiles may also partly be explained by deviating peak temperatures of the solar cells $T_W$ during contact firing. It has been reported in [164] that due to higher absorbance of the higher doped boron emitter profiles (lower $R_{sh}$), $T_W$ of these solar cells can be higher compared to the solar cells with boron emitter with lower doping (higher $R_{sh}$). Although this effect cannot be excluded completely, it is assumed that it only plays a minor role in this investigation, since the absolute difference between the $j_{02}$ values for the different boron emitter profiles is higher than the change of $j_{02}$ for a wide variation of $T_S$ (for $T_S \leq 880^\circ C$), as depicted in Figure 3.30. Moreover, a boron emitter with lower $R_{sh}$ might also have a higher defect concentration in the SCR than a boron emitter with higher $R_{sh}$, due to defective incorporation of boron atoms into the Si crystal lattice for high boron concentrations. This might also support parasitic leakage currents, which potentially explains a part of the higher $j_{02}$ values observed for the boron emitters with lower $R_{sh}$. In addition, the strongly increasing $j_{02}$ for all boron emitter for $T_S \geq 880^\circ C$ might also partly be explained by diffusion of paste components to the SCR, as observed in [169] for pure Ag pastes. Although Ag/Al pastes are based on pure Ag pastes with an addition of around 5% Al and diffusion of paste components to the SCR cannot be excluded, its influence on $j_{02}$ of the solar cells is expected to play a minor role. Possibly the boron emitter profile also affects the mechanism of Ag/Al spike formation slightly. Because of the strong variation of the depth of the Ag/Al metal spikes, observed by the SEM analysis (see Figure 3.25), it was not possible to clarify this issue.

An excellent method for visualizing the microscopic Ag/Al spikes and the thereby induced leakage currents is the dark lock-in thermography (LIT) measurement. By applying an alternating voltage $V$ to the solar cell using the LIT technology, the areas affected by parasitic leakage currents due to the Ag/Al metal spikes heat up by ohmic power dissipation. The heat signals can be measured by an IR camera. Figure 3.32 exemplarily shows the LIT measurement for two bifacial n-PERT solar cells fired with $T_S = 830^\circ C$ (left side), respectively $T_S = 920^\circ C$ (right side). It is evident, that the amount and size of the heat signals increase substantially for increasing $T_S$. These findings agree well with the observation of the Ag/Al formation presented in section 3.4.3.
3 Metallization-Induced Power Loss of Bifacial n-Type Silicon Solar Cells

Figure 3.32: Dark lock-in thermography measurements ($V \approx 0.5$ V for $\approx 30$ min) of bifacial solar cells fired with set peak firing temperature $T_S = 800^\circ$C (left side) and $T_S = 920^\circ$C (right side). The thermal signals correlate to leakage currents induced by Ag/Al metal spikes penetrating the SCR. The amount and size of the heat signals increases substantially for increasing $T_S$.

3.4.5 Optimization of rear side metallization grid

Based on the findings on contact formation on boron emitters using screen-printing of Ag/Al metal pastes and subsequent contact firing, presented in the previous sections, the ideal Ag/Al H-grid metallization is computed. Similar to the computation of the ideal H-grid metallization of screen-printed Ag metal contacts on the phosphorus-doped FSF, presented in section 3.3.5, the whole IV-curve is computed according to the two-diode model for a wide variation of the finger width $w_F$ and distance between adjacent fingers $d_F$. From these computed IV-curves the relevant IV-characteristics $\eta$, $j_{sc}$, $V_{oc}$ and $FF$ are extracted.

It was demonstrated in the previous section that the metallization-induced power loss increases considerably for increasing $T_S$. The bifacial n-PERT solar cells with highest $\eta$ were achieved for $T_S = 760^\circ$C for all of the studied boron emitter profiles (see Figure 3.26a). Since $T_S = 760^\circ$C is close to the temperature range of inappropriate contact formation (see strong increase of $R_S$ for $T_S < 760$ C in Figure 3.26a), $T_S = 800^\circ$C was used for the simulation. Moreover, the power loss due to recombination in the SCR induced by the Ag/Al metallization turned out to be lower on boron emitters with higher $R_{sh}$. Therefore, the boron emitter profile with $R_{sh} = 64$ $\Omega$/sq and Al$_2$O$_3$/SiN$_x$ passivation layer stack was used for the simulation, although Al$_2$O$_3$/SiN$_x$ degrades considerably for increasing $T_S$. The measured $\rho_c \approx 3.5$ m$\Omega$cm$^2$ was implemented for the used Ag/Al paste P6 at $T_S = 800^\circ$C (see Figure 3.19b). $\rho_c$ of the screen-printed Ag/Al metal contacts, as a function of the width of the screen-opening $w_S$, was determined by fitting the function $f(w_S) = a/w_S^b$ to the measurement data, illustrated in Figure 3.33. Note that frequently used screens cause strong broadening of the metal fingers of around 25 – 40 $\mu$m. For new screens, the broadening of the metal fingers is only in the range of 5 – 15 $\mu$m, as shown in section 3.3.2. A broadening of the fingers of 10 $\mu$m was assumed for the simulation.
To include the power losses caused by recombination at the metal contacts of the solar cells, the emitter saturation current densities of the Ag paste \(j_{01c,Ag}\) and Ag/Al paste \(j_{01c,Ag/Al}\) must be calculated. The values for the Ag metal contacts were taken from literature, since they are well known. With the metallization fractions of the Ag metal contacts \(f_{c,Ag}\) and the one of the Ag/Al metal contacts \(f_{c,Ag/Al}\), the saturation current density of the first diode \(j_{01c,Ag/Al}\) induced by the Ag/Al metal contacts can be calculated by

\[
j_{01c,Ag/Al} = \left( j_{01c} - f_{c,Ag} \cdot j_{01c,Ag} \right) / f_{c,Ag/Al}.
\]

With the total saturation current density of the first diode caused by the front and rear metal contacts \(j_{01c} = 216.7\) fA/cm\(^2\) (see Figure 3.29d), \(j_{01c,Ag} = 500\) fA/cm\(^2\) on the phosphorus FSF with \(R_{sh} = 60\) Ω/sq (published in [166]), \(f_{c,Ag} = 0.069\) (\(d_F = 2\) mm, \(w_F \approx 100\) µm, \(w_{BB} = 1.4\) mm) and \(f_{c,Ag/Al} = 0.064\) (\(d_F = 2.35\) mm, \(w_F \approx 100\) µm, \(w_{BB} = 1.5\) mm), the saturation current density of the Ag/Al metal contacts calculates to \(j_{01c,Ag/Al} = 2.855.8\) fA/cm\(^2\). This high value fits well to the values reported in literature [140, 165]. A similar calculation can be performed for the saturation current density of the second diode caused by the Ag/Al metal contacts

\[
j_{02c,Ag/Al} = \left( j_{02c} - f_{c,Ag} \cdot j_{02c,Ag} \right) / f_{c,Ag/Al}.
\]

With the total saturation current density of the second diode due to the front and rear metal contacts \(j_{02c} = 45\) nA/cm\(^2\) (see Figure 3.30b) and \(j_{02c,Ag} = 20\) nA/cm\(^2\) on the phosphorus FSF with \(R_{sh} = 60\) Ω/sq (reported in [86]), the saturation current density of the Ag/Al metal contacts calculates to \(j_{02c,Ag/Al} = 683.7\) nA/cm\(^2\). These calculations prove the much higher recombination activity of the Ag/Al metal contacts compared to pure Ag metal contacts.

The implemented contribution of the rear side to the total \(R_S(w_F, d_F)\) of the solar cells consists of the contact resistance of the Ag/Al-Si interface \(R_c(w_F, d_F)\) (equation (3.25)), lateral resistance between the Ag/Al metal fingers in the boron emitter \(R_e(w_F, d_F)\) (see equation (3.24)) and line resistance of the Ag/Al metal fingers \(R_l(w_F, d_F)\) (equation (3.17)). Note that the non-simplified equations were used each. \(r_i(w_F, d_F)\) was implemented by the fit function \(f(w_S) = 25,129 / w_S^{1.6}\).
(see Figure 3.33) assuming $w_S = w_F - 10\ \mu m$. For the computation, a s-FSF was implemented instead of a homogeneous FSF to achieve higher $\eta$ of the solar cells. All parameters of the front side Ag metallization on the s-FSF were kept constant, based on the computation of the ideal Ag H-grid presented in section 3.3.5 ($w_F = 70\ \mu m$, $d_F = 2\ mm$ and $w_{BB} = 0.6\ mm$), since they do not affect the characteristics of the rear side metallization. The implemented contribution of the front side to the total $R_S$ comprises the lateral resistance of the FSF ($R_{FSF} = 0.15\ \Omega cm^2$), contact resistance of the Ag metal contacts ($R_{c,f} = 0.014\ \Omega cm^2$) and line resistance of the Ag metal fingers ($R_{lf} = 0.15\ \Omega cm^2$).

Power losses caused by recombination at the front and rear side metallization were implemented according to equation (3.28), using the $j_{01c}$ and $j_{02c}$ values presented above. For the passivated area of the boron emitter on the rear side, $j_{01e} = 44\ fA/cm^2$ (see Table 1 in section 2.5.4) was used, whereas for the s-FSF on the front side, $j_{01FSF} = 85\ fA/cm^2$ (outlined in section 3.3.4) was implemented. An overview of the implemented values used for the computation of the IV-curves of 6-inch bifacial RJ n-PERT Si solar cells with homogeneous boron emitter and s-FSF for different rear side metallization parameters is presented in Table 5.

<table>
<thead>
<tr>
<th>$l_S$ [mm]</th>
<th>$l_D$ [mm]</th>
<th>$l_F$ [mm]</th>
<th>$N_{BB}$</th>
<th>$w_{BB}$ [mm]</th>
<th>$R_{FSF}$ [$\Omega cm^2$]</th>
<th>$R_{c,f}$ [$\Omega cm^2$]</th>
<th>$R_{lf}$ [$\Omega cm^2$]</th>
</tr>
</thead>
<tbody>
<tr>
<td>156</td>
<td>1.0</td>
<td>25.37</td>
<td>3</td>
<td>0.6</td>
<td>0.15</td>
<td>0.014</td>
<td>0.15</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>$\rho_c$ [m$\Omega cm^2$]</th>
<th>$r_i$ [m$\Omega /mm$]</th>
<th>$j_{01c,Ag}$ [fA/cm²]</th>
<th>$j_{01c,Ag/Al}$ [fA/cm²]</th>
<th>$j_{02c,Ag}$ [fA/cm²]</th>
<th>$j_{02c,Ag/Al}$ [fA/cm²]</th>
<th>$j_{01e}$ [fA/cm²]</th>
<th>$j_{01FSF}$ [fA/cm²]</th>
</tr>
</thead>
<tbody>
<tr>
<td>3.5</td>
<td>25,129/$w_S$\textsuperscript{1.6}</td>
<td>500</td>
<td>2,855.8</td>
<td>20</td>
<td>683.7</td>
<td>44</td>
<td>85</td>
</tr>
</tbody>
</table>

Figure 3.34 shows the IV-characteristics, extracted from the computed IV-curves of the solar cells, for a wide range of $w_F$ (20 - 140\ \mu m) and $d_F$ (1 - 3\ mm). Figure 3.34a shows that $FF$ of the solar cell is not continuously increasing for increasing metallization fraction $f_c$ of the Ag/Al metal contacts, although $R_S$ is a continuously decreasing function of $f_c$ (see Figure 3.34b). This can be explained by an increasing $j_{02}$ of the solar cells for increasing $f_c$, which is induced by the increasing area density of the Ag/Al metal spikes (see section 3.4.3). The enhanced $j_{02}$ results in a loss of $FF$, overcompensating the effect of the decreasing $R_S$. Although the parameters for the lowest recombination activity of the Ag/Al metal spikes were used for the computation, their influence on $FF$ has to be considered. As expected, $V_{oc}$ is continuously decreasing for increasing $f_c$, showing a much more distinct decay than the one observed for the Ag metal contacts (compare Figure 3.34c with Figure 3.15c). Although no power loss due to shading of the rear side by the Ag/Al metal contacts is considered in this computation, it turns out that the
optimal parameters, in terms of a high $\eta$, of the Ag/Al H-grid ($w_F \approx 60 \, \mu m; \, d_F = 1.6 - 2.0 \, mm$) are similar to the ones observed for the Ag H-grid on the front side, where power loss due to shading of the wafer surface by the Ag contacts is considered. This is evident by comparing Figure 3.34d with Figure 3.15d, indicating that the power loss due to recombination at the Ag/Al metal contact is in the same range as the power loss due to shading and shows a similar behavior for increasing $f_c$.

**Figure 3.34:** IV-characteristics $FF$, $R_S$, $V_{oc}$ and $\eta$ of bifacial RJ n-PERT Si solar cells with *co-diffused* boron emitter and phosphorus s-FSF, extracted from the computed IV-curves (two-diode model), as a function of finger width $w_F$ and finger spacing $d_F$ of the Ag/Al rear side metallization H-grid. Note the non-linear scaling of $FF$, $R_S$ and $\eta$.

### 3.5 Optimized solar cells and loss analysis

Based on the findings on contact formation via screen-printing of Ag/Al pastes, bifacial n-PERT Si solar cells were manufactured according to the process sequence described in section 3.1. Optimized process parameters such as an optimized *co-diffusion* were conducted, targeting high $\eta$ of the solar cells. As base material, n-type Cz-Si (6 - 8 $\Omega cm$) was utilized and the wafers were
alkaline textured on both sides. Boron emitter and phosphorus FSF were formed in one single high temperature step in a POCl₃ tube diffusion furnace using a two-step process according to the one described in section 2.5.1. The s-FSF was formed by inkjet-printing of an etch-resistant mask and selective EB from \( R_{sh} \approx 35 \Omega/sq \) to \( R_{sh} \approx 70 \Omega/sq \). The sheet resistance of the boron emitter was experimentally determined by 4PP measurement revealing \( R_{sh} \approx 70 \Omega/sq \). Contact formation was performed by screen-printing of the commercial Ag paste P1 on the phosphorus s-FSF and the commercial Ag/Al paste P5 on the boron emitter. Since contact firing is one of the key processes and to prove the observations of the previous section on the ideal \( T_S \) again, the solar cells were fired in an IR conveyor belt furnace with \( T_S \) in the range from 750 – 840°C. The solar cells were characterized using a flasher device from the manufacturer h.a.l.m, which is equipped with an absorbing black back sheet.

Figure 3.35 illustrates the solar cell IV-characteristics of the large area (156.25 cm²) bifacial n-PERT solar cells, averaged over four solar cells. Since the solar cells are bifacial, their IV-characteristics can be measured from both sides. The primary solar cell process sequence was configured for RJ cells. However, it turned out in this experiment that the best \( \eta \) can be achieved for the front junction FJ configuration, even when using a Ag/Al metallization pattern optimized for RJ n-PERT cells. Best results are obtained for the solar cells fired with \( T_S \) in the wide range of 780 – 840°C. At \( T_S = 750°C \) the fill factor \( FF \) drops strongly because of insufficient contact formation of the Ag/Al metal paste on the boron emitter. As expected from the previous studies, \( j_{sc} \) is nearly independent of \( T_S \). \( V_{oc} \) is only slightly reduced for increasing \( T_S \) caused by a minimal degradation of the Al₂O₃/SiNx:H passivation layer stack and due to an increasing \( j_{01} \) induced by the metallization. Indeed, a stronger degradation of the Al₂O₃/SiNx:H passivation layer stack was expected from the previous experiments. Since the rear side Ag/Al metallization fraction was significantly lower in this than in the previous study, the compromising influence of the Ag/Al metal spikes, resulting in a decrease of \( FF \) for increasing \( T_S \), is less pronounced in this temperature range. Therefore, the \( FF \) curve is continuously increasing due to a continuously decreasing \( R_S \) for increasing \( T_S \).

Table 6 shows the IV-characteristics of the best bifacial n-PERT Si solar cell for ideal \( T_S \). To estimate the gain of the bifaciality, the solar cells were measured with a flasher tool with an absorbing black back sheet and additionally with a reflective white back sheet. Best results were obtained under FJ illumination conditions. This indicates that the saturation current density of the boron emitter (passivated with the Al₂O₃/SiNx:H passivation layer stack) is lower than the one of the phosphorus FSF/BSF (passivated with the SiO₂/SiNx:H passivation layer stack). This assumption is confirmed by measurements of the internal and external quantum efficiency (IQE/EQE) shown in Figure 3.36. The best cell achieved a conversion efficiency of 19.7% with \( V_{oc} = 653 \) mV under FJ illumination, when applying the white back sheet. The measured \( V_{oc} \) values on the solar cells of above 650 mV show the high electrical quality of the boron emitters by means of co-diffusion from a boron-doped doping source fabricated by PECVD. Moreover, the values are equal or even higher than \( V_{oc} \) values obtained on bifacial n-PERT Si solar cells.
with boron emitters from common BBr₃-diffusion, as reported in [133, 170-172]. This shows the industrial feasibility of co-diffusion for boron emitter formation. Rothhardt et al. reported \( \eta = 19.9\% \) on co-diffused bi-facial n-PERT solar cells using APCVD [58]. This high value can mainly be explained by an advanced metallization process using dual print on both wafer sides, compared to a single screen-printing step used in this work. Cabal et al. showed \( \eta = 20.3\% \) on co-diffused bi-facial n-PERT solar cells using the doping layers multifunctional [173]. These outstanding results may also partly be explained by an advanced metallization process.

![Figure 3.35: IV-characteristics of bifacial n-PERT solar cells fired at low set peak firing temperatures \( T_s \) (averaged over four solar cells each). Best cell results are obtained for \( T_s \) in the range of 780 – 840°C. The data was measured with a non-reflecting black back sheet. The lines serve as guides to the eye. Published in [92].](image)

The bifacial n-PERT Si solar cells presented above suffer mainly from low FF. This can mainly be attributed to a high \( R_s \) of about 1.0 – 1.3 \( \Omega \text{cm}^2 \). Therefore, a loss analysis of the different contributions of \( R_s \) was performed. Figure 3.37 depicts the different components contributing to the ohmic power loss induced by \( R_s \). The calculations are based on the measurement data. The line resistance \( R_l \) of the screen-printed Ag/Al metal fingers, calculated by equation (3.17), was identified as main contributor to \( R_s \) with a portion of almost 46%. This can be explained
by the high line resistivity of 70 mΩ/mm, which was experimentally determined. $R_l$ of the screen-printed Ag fingers exhibits an essentially lower contribution of 15.4% to $R_S$.

**Table 6:** Results of the best bifacial n-PERT Si solar cells with optimized PECVD boron emitter and set peak firing temperature $T_S$. The data was measured with a flasher tool with an absorbing black and a reflective white back sheet.

<table>
<thead>
<tr>
<th></th>
<th>$V_{oc}$ [mV]</th>
<th>$j_{sc}$ [mA/cm$^2$]</th>
<th>FF [%]</th>
<th>$\eta$ [%]</th>
<th>$R_S$ [Ωcm$^2$]</th>
</tr>
</thead>
<tbody>
<tr>
<td>black back sheet (rear junction)</td>
<td>652</td>
<td>37.9</td>
<td>76.3</td>
<td>18.8</td>
<td>1.1</td>
</tr>
<tr>
<td>black back sheet (front junction)</td>
<td>652</td>
<td>38.9</td>
<td>76.3</td>
<td>19.3</td>
<td>1.0</td>
</tr>
<tr>
<td>white back sheet (front junction)</td>
<td>653</td>
<td>39.4</td>
<td>76.4</td>
<td>19.7</td>
<td>1.0</td>
</tr>
</tbody>
</table>

**Figure 3.36:** Measurement of the internal/external quantum efficiency (IQE/EQE) of the best bifacial n-PERT solar cell. The IQE is higher for front junction illumination. Published in [92].

Furthermore, it was calculated, corresponding to equation (3.25), that the contact resistance $R_c$ of the Ag/Al metal fingers on the boron emitter is in charge of 18.3% of the total $R_S$. In contrast, $R_c$ of the screen-printed Ag metal fingers on the phosphorus-doped FSF has only a small contribution of 2.2% to $R_S$. The contribution of the lateral resistance of the FSF and emitter can be quantified by equation (3.24) to be around 6.2% and 9.4% respectively. The resistance of the base shows also a contribution of 2.7%. The calculated contributions, based on experimental data, cumulate to $R_S = 1.05$ Ωcm$^2$, which is in good agreement with the measured value of $R_S$. Note that the metallization is responsible for almost 82% of the total $R_S$ of the bifacial n-PERT solar cells. The Ag/Al metallization is responsible for over 64% of $R_S$ and therefore the major source for the limitation of $FF$. 
Figure 3.37: Pie chart of the different components contributing to the total ohmic losses induced by the series resistance $R_s$ of the n-type bifacial PERT solar cells.

3.6 Summary

In this chapter, the screen-printing technology and its practical application for the metallization of bifacial n-PERT solar cells was outlined. It was demonstrated, that the line resistance of screen-printed metal fingers depends strongly on the paste composition, the parameters of the mesh and screen-printing parameters like the printing velocity. Best results for narrow Ag metal fingers were obtained for the mesh with the parameter 400/18 at low printing speed of 50 mm/s. Furthermore, it was demonstrated that the specific contact resistance $\rho_c$ of the screen-printed Ag contacts on the highly-doped regions of the selective phosphorus FSF with $R_{sh} = 40 \ \Omega/$sq is below 1 m$\Omega$cm$^2$, being almost independent of the set peak firing temperature $T_S$. Therefore, the contribution of $\rho_c$ of the Ag metal contacts to the total power loss of the n-PERT Si solar cells can be neglected. Based on the measurements of the line and contact resistance, an optimization of the front side metal pattern was presented under consideration of all relevant power loss mechanisms that are related to the metallization. The computation shows that the highest $\eta$ can be achieved for thin fingers with $w_F \approx 40 \ \mu$m and $d_F = 1.4 – 1.9$ mm. However, narrowing of the metal fingers due to technical issues during screen-printing, results in a strong decrease of $\eta$ in this region of $w_F$. A much more insusceptible region with comparable values in $\eta$ was found for screen-printed Ag metal fingers with $w_F \approx 60 \ \mu$m and $d_F = 2.0 – 2.2$ mm, leading to a higher process stability. Moreover, it was shown that recombination losses due to the Ag metallization are low compared to power losses due to resistance and shading.

Thereafter, the limiting rear side metallization by means of screen-printing of Ag/Al pastes for the contact formation on boron-doped emitters was discussed. It was shown that $\rho_c$ of the Ag/Al
metal contacts depends strongly on $T_S$ and exhibits significantly higher values ($\rho_c \approx 4 \text{ m}\Omega\text{cm}^2$) on boron emitters with $R_{sh} = 70 \text{ }\Omega/$sq at $T_S = 800^\circ\text{C}$, compared to the Ag metal contacts. For lowly doped boron emitters ($R_{sh} > 100 \text{ }\Omega/$sq), the contact resistance can be limiting for the fill factor of the solar cells ($\rho_c > 10 \text{ m}\Omega\text{cm}^2$). In a detailed microstructural study on the formation of Ag/Al crystallites during contact firing it was shown that the area density of the metal spikes increases strongly for increasing $T_S$. This was verified using top-view SEM images of etched contact spots. Moreover, it could be experimentally determined that the depth of the Ag/Al metal spikes is above 1 $\mu$m even for low $T_S$. Therefore, the metal spikes penetrate the SCR, inducing a high saturation current density of the second diode $j_{02}$. A compromise in the firing conditions between appropriate contact formation in terms of low $\rho_c$ and low amount of Ag/Al spikes has to be made. The total $V_{oc}$ loss induced by the metallization was calculated to be in the range of 20 - 90 mV, being almost independent of the applied passivation layer stack.

Furthermore, it was demonstrated that the recombination activity of the Ag/Al metal contacts is much higher than the one of pure Ag metal contacts. The total saturation current density of the first diode caused by the front and rear metal contacts was calculated to around 217 fA/cm$^2$ at $T_S = 800^\circ\text{C}$ for metallization fractions of around 6 – 7% on both sides. From this value, the saturation current density of the Ag/Al metal contacts calculates to about 2856 fA/cm$^2$. A similar calculation revealed a saturation current density of the second diode of the Ag/Al metal contacts of about 684 nA/cm$^2$. Furthermore, it turned out that the detrimental influence of the Ag/Al spikes on $FF$, $V_{oc}$ and therefore on $\eta$ is more distinct for solar cells with boron emitters with lower $R_{sh}$, even though these emitters show a better electrical shielding of the physical metal contacts, strongly increasing for increasing $T_S$. Although the doping profiles for boron emitter with lower $R_{sh}$ are deeper, the Ag/Al metal spikes are deep enough (> 1 $\mu$m) to easily penetrate the SCR, even for low $T_S$. The Ag/Al spikes represent direct contacts between the highly-doped boron emitter and the metal (Schottky-contacts). The width of the Schottky barrier is lower for a higher doping of the boron emitter. This results in a higher probability for charge carriers to tunnel through the Schottky barrier and contribute to parasitic leakage currents in the case of a Ag/Al metal spike breaking through a highly-doped boron emitter. Therefore, it is assumed that the detrimental leakage currents are proportional to the doping concentration and also explain the observed lower $\rho_c$ of the boron emitters with lower $R_{sh}$. These findings agree with existing analytical models for the calculation of the specific contact resistance presented by Lohmüller $et$ $al.$ in [163] and Fritz $et$ $al.$ in [157], explaining the experimentally observed lower contact resistance for boron emitters with lower $R_{sh}$. With these findings, the ideal Ag/Al metallization H-grid was computed for the lowly-doped boron emitter with $R_{sh} = 64 \text{ }\Omega/$sq and Al$_2$O$_3$/SiNx passivation layer stack at $T_S = 800^\circ\text{C}$. The simulation revealed similar optimal parameters of the Ag/Al H-grid ($wF \approx 60 \mu$m and $dF = 1.6 – 2.0 \text{ mm}$) as observed for the Ag H-grid on the front side, indicating that the power losses due to recombination at the Ag/Al metal contact is in the same range as the power losses due to shading.
The industrial feasibility of the alternative method for boron emitter formation by means of co-diffusion from a doped PECVD layer was demonstrated on large area (156.25 cm²) bifacial n-PERT Si solar cells with conversion efficiencies of up to 19.7%. The measured $V_{oc}$ values on the solar cells of above 650 mV show the high electrical quality of the co-diffused boron-doped emitters. A loss analysis revealed that the metallization is responsible for almost 82% of the total series resistance $R_S$ of the bifacial n-type PERT Si solar cells presented in this work. The line resistance of the screen-printed Ag/Al metal finger was identified to be the main contribution to $R_S$ with a portion of almost 46%. The contribution of the contact resistance of the Ag/Al metal fingers to $R_S$ was calculated to be 18.3%. This demonstrated that the Ag/Al metallization is the major source for the limitation of $FF$.

The presented results of the bifacial n-PERT solar cells ($\eta = 19.7\%$) are comparable to the ones published by other research groups. Rothhardt et al. reported $\eta = 19.9\%$ on co-diffused bi-facial n-PERT solar cells using APCVD [58]. This higher $\eta$ value can mainly be explained by an advanced metallization process using dual print on both wafer sides, compared to the single screen-printing step used in this work. Cabal et al. showed $\eta = 20.3\%$ on co-diffused bi-facial n-PERT solar cells using the doping layers multifunctional [173]. These outstanding results may also partly be explained by an advanced metallization process.
3 Metallization-Induced Power Loss of Bifacial n-Type Silicon Solar Cells
4 Mono-Facial n-Type Silicon Solar Cells with LCO

As demonstrated in the previous chapter, screen-printing metallization on boron emitters using Ag/Al pastes results in essential power loss due to enhanced series resistance and recombination at the metal contacts as well as in the SCR, limiting the efficiency of bifacial n-PERT Si solar cells. In this chapter, an alternative approach for contact formation on boron emitters, by means of local contact opening (LCO) using laser ablation of the rear side passivation layer stack and subsequent full-area deposition of Al, is studied on mono-facial n-PERT Si solar cells. This approach allows to substantially reduce the metallization fraction on the boron emitter, which noticeably lowers recombination at the metal contacts. Moreover, contact firing is not required for contacting the boron emitter, obviating metal spike formation. Although this approach has a higher efficiency potential compared to screen-printing of Ag/Al pastes, the fill factor \( FF \) of mono-facial n-PERT solar cells with LCO and evaporated Al can be lower than expected, as reported e.g. in [19]. This \( FF \) loss has been attributed to an increased saturation current density of the second diode \( j_{02} \), potentially induced by the local contact formation, without clarifying its physical origin. Therefore, this chapter is focused on giving the missing explanation of the loss mechanisms causing the limitation of \( FF \). An empirical model for the distribution of recombination centers (RCs), potentially induced by LCO and subsequent evaporation of Al, is presented. This model is validated by three-dimensional (3D) numerical simulation, describing the detrimental influence of the RCs on the IV-characteristics of mono-facial n-PERT solar cells and thereby successfully reproducing the measured IV-characteristics. Furthermore, the optimal contact geometry of the LCO is simulated to estimate the efficiency potential of mono-facial n-PERT solar cells with \textit{co-diffused} boron emitters. Finally, a detailed free energy loss analysis is presented to get a deeper understanding of the relevant loss mechanisms.

4.1 Mono-facial n-PERT solar cell process

In this section, the mono-facial RJ n-PERT Si solar cell process, which was used in this work, is presented. Figure 4.1 depicts the process flow, which is similar to that for bifacial n-PERT solar cells, as described in section 3.1, except for the structuring, passivation and metallization of the rear side. First, the wafers are chemically polished in an aqueous solution of KOH to remove the saw damage (removal \( \approx 15 \) \( \mu \)m per side). Subsequently, the rear side is capped with a SiNx:H layer by means of PECVD. This enables only the front surface to be structured in the subsequent alkaline texturing step, because the rear surface is protected by the SiNx:H capping layer. While the thickness of the capping layer is not critical, since SiNx:H is not affected by KOH, it must be ensured that the capping layer has only a low density of pinholes. Moreover, care must be taken that the capping does not wrap around to the front surface during deposition. However, this wrap-around can be removed by a short etch in HF (2\%). It turned out that this process sequence has advantages in terms of a low wafer roughness, compared to the alternative procedure, namely, initial texturing and subsequent polishing of the wafers. This is based on
the observation that it is hard to adequately remove the random pyramids with the KOH-based polishing solution. The enlarged wafer surface for rough surfaces may result in disadvantages in terms of passivation quality and rear reflectivity for the long-wave light. Furthermore, the laser ablation process may be compromised on a rough surface, as discussed in section 4.2.1. After texturing the front side, the capping layer is removed in HF (2 – 10%) and the wafers are cleaned by a wet chemical piranha cleaning to remove metallic and organic residues from the wafer surface. The wafers have to be transported in an ultra-clean box to avoid contamination before the deposition of the SiO$_x$:B doping layer. The doping layer is deposited on the rear side of the wafers via PECVD and is capped with a SiN$_x$:H capping layer with a thickness of 120 nm, which can be deposited in the same PECVD device. Boron emitter and phosphorus FSF are formed in sequential co-diffusion in a POCl$_3$ quartz tube diffusion furnace, corresponding to the procedure described in section 2.5.2. Following that, the doping and capping layers, as well as the PSG, are removed in an aqueous solution of HF (10%).

The formation of the selective FSF is realized by inkjet-printing of an etch-resistant ink and EB of the highly-doped FSF, according to the procedure described in section 2.5.5. After EB, the wafers are again cleaned in the piranha cleaning solution to remove potential residues from the selective EB. Surface passivation of the s-FSF is established by thermal oxidation at 900°C for 5 minutes, resulting in a SiO$_2$ layer with a thickness of about 7 nm. Since the SiO$_2$ layer must be removed on the rear side before deposition of the Al$_2$O$_3$ layer for the passivation of the boron emitter, an SiN$_x$:H layer with a thickness of 87 nm thick is deposited on the front side by means of PECVD. The thickness was calculated in accordance to the fact that during removal of the SiO$_2$ layer on the rear side and additional cleaning step, the SiN$_x$:H layer on the front side is also etched. This was discussed in more detail in section 2.5.5. After removal of the SiO$_2$ layer and piranha cleaning, the thickness of the SiN$_x$:H layer on the front side can be estimated to be 67 nm, which is ideal for ARC. Thereafter, an Al$_2$O$_3$ layer with a thickness of 10 nm is deposited on the rear side by PA-ALD, enabling excellent surface passivation on boron emitters, as shown in section 2.6.2. The Al$_2$O$_3$ is capped by a SiN$_x$:H layer with a thickness of 120 nm.

**Figure 4.1:** Process flow of the mono-facial n-type PERT solar cell process.

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by PECVD, in comparison to the bifacial solar cell process, where a SiNₓ:H layer with a thickness of 65 nm is applied on the rear side. The thicker SiNₓ:H layer on the flat rear side improves the rear reflectivity of the solar cell. Contact formation on the boron emitter via screen-printing of commercial Ag/Al pastes would lead to an excessively high specific contact resistance for such a thick SiNₓ:H layer, as discussed in section 3.4.1. Therefore, an alternative method for contact formation is used, as discussed in the following.

Contact formation on the s-FSF is conducted by means of screen-printing of a commercial Ag paste (see section 3.3). Subsequently, the contacts are fired in an IR belt conveyor furnace with set peak firing temperature $T_\text{S} = 800^\circ\text{C}$ at a belt speed of 6 m/min, enabling Ag metal contacts on the front side with a specific contact resistance $\rho_c < 1 \text{ m}\Omega\text{cm}^2$ (see section 3.3.3), while the passivation quality of the Al₂O₃/SiNₓ:H passivation layer stack is hardly reduced for this low $T_\text{S}$ (see section 3.4.4). After contact firing, the contact formation on the boron emitter is realized by local contact formation, as described in the following sections. Figure 4.2 illustrates the schematic cross section of a mono-facial n-PERT Si solar cell with co-diffused boron emitter.

![Schematic cross-section of a mono-facial n-PERT solar cell with co-diffused boron emitter](image)

**Figure 4.2:** Schematic cross-section of a mono-facial n-PERT solar cell with co-diffused boron emitter and phosphorus FSF. The contact formation on the FSF is conducted via screen-printing of a commercial Ag paste and firing in an IR conveyor belt furnace. Contact formation on the boron emitter is realized by LCO and subsequent deposition of a 1 µm thick Al layer. The boron emitter is passivated by an Al₂O₃/SiNₓ:H layer stack, whereas the FSF is passivated by a SiO₂/SiNₓ:H layer stack.

### 4.2 Local contact formation

For industrially manufactured bifacial n-PERT solar cells, usually screen-printing of Ag/Al pastes is utilized for contact formation on the boron emitter (see section 3.4). As demonstrated in section 3.4.4, this leads to a total open circuit voltage $V_{oc}$ loss of around 30 mV even for ideal contact firing conditions. This $V_{oc}$ loss can mainly be attributed to the Ag/Al metal contacts, which induce a high saturation current density of the first diode $j_{01}$. In addition, the Ag/Al metal spikes penetrate through the SCR, causing a high saturation current density of the second diode $j_{02}$, which mainly limits the fill factor $FF$ of the solar cells (see section 3.4.3).

To avoid power loss attributed to the Ag/Al metallization on the boron emitter and to improve the internal rear reflectivity, an alternative metallization technique for local contact formation

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is used for the mono-facial n-PERT solar cells. By this means, the dielectric passivation layer stack on the rear side is locally removed by laser ablation. Subsequently, a full-area Al contact with a thickness of 1 µm is deposited on top of the passivation layer stack and forms local contacts to the boron emitter in the regions where the passivation layer has been removed. For the laser ablation process, different methods for the deposition of the full-area Al contact and the influence of different contact geometries on the characteristics of mono-facial n-PERT solar cells are discussed in the following subsections.

4.2.1 Laser ablation

First, the dielectric passivation layer stack on the boron emitter must be locally removed [174, 175]. This was performed in this work by laser ablation. Since Gaussian-shaped laser pulses were exclusively used in this work, some of the following explanations apply for Gaussian-shaped laser pulses only. The laser ablation process depends essentially on the pulse length and wavelength of the laser source. The use of a laser source with pulse length in the range of nanoseconds (ns) results in substantial thermal damaging of the Si substrate underneath the dielectric layers [176]. If the pulse length of the laser source is in the range of femtoseconds (fs), high energies are required to reach the ablation threshold, which is defined as the minimal fluence required for complete removal of the dielectric layers (the fluence can be calculated by the peak energy and the width of the laser spot). Thus, a significant portion of energy is released in the Si substrate underneath without sufficient ablation of the dielectric layers [177]. By this means, recombinative defects can be generated in the Si substrate underneath the passivation layer. These RCs compromise the IV-characteristics of Si solar cells (see section 2.3.4) and must be avoided to achieve a high conversion efficiency.

A compromise is the use of a laser source with a pulse length in the range of picoseconds (ps). The absorption mechanisms are mainly linear and the thermal damaging can be reduced, since the dielectric layer is not removed by heating of the material underneath but mainly by direct transition into its plasma state [178]. The energy is primarily dissipated in kinetic energy of the plasma expansion. Therefore, the ablation process using ps-pulses is also referred to as cold ablation. In this work, a laser system from the manufacturer 3D-Micromac was used, which is equipped with the laser source Talisker™ from the manufacturer Coherent. The laser system enables the emission of laser light with the wavelengths 1064 nm, 532 nm and 355 nm. The pulse length of the Gaussian-shaped laser pulses is in the range of 10 -15 ps, depending on the used wavelength. The predominant mechanism for the wavelengths 1064 nm and 532 nm is indirect ablation, since the dielectric passivation layers used in this work, such as Al2O3, SiO2 and SiNx:H, are almost completely optically transparent for these wavelengths [179]. Figure 4.3 illustrates the mechanism of indirect ablation, which can be explained by blistering of the dielectric layers due to the gas pressure of the evaporated Si substrate and plasma underneath. For a certain laser pulse length, the laser ablation process depends primarily on the fluence of the laser pulse. The laser spot size can be adjusted by defocusing. The laser power is adjusted to a
value which only slightly exceeds the ablation threshold. Thus, melting of the material beneath the laser spot, causing a thermal shock wave in the crystal lattice which potentially induces defects, can be minimized. Optical microscopy of the laser spots allows to examine, if the laser parameters are adjusted ideally, for sufficient ablation of the dielectric layer. For plane wafer surfaces, laser ablation results in a circular ablation edge when using Gaussian-shaped laser pulses, as reported in [178]. If laser ablation is applied on alkaline textured surfaces, the ablation mechanism is different because of optical interference at the flanks of the pyramids. This may lead to insufficient ablation of the dielectric layer at the top of the pyramids, as reported in [178] using SiNₓ:H layers with a thickness of 75 nm, which are comparable to the passivation layers used in this work. In addition, laser-induced damage is potentially enhanced in the areas of constructive interference. For this reason, laser ablation was exclusively conducted on flat wafer surfaces in this work.

Considering the Al₂O₃/SiNₓ:H dielectric layer stack utilized in this work for the passivation of the boron emitter (see section 2.6.2), the laser ablation process was found to be optimal for ps-pulses with a wavelength of 532 nm [60]. Because of the high doping of the boron emitter, FCA plays an essential role at the wafer surface, which leads to an increased absorption of photons close to the wafer surface [180]. Note that FCA is negligible in the Si base substrate, as discussed in section 2.3.1. In addition, non-linear absorption mechanisms become relevant [181]. By this means, the ablation fluence threshold is reduced, leading to less damage in the Si substrate. It turned out that the minimal laser fluence for sufficient ablation of the Al₂O₃/SiNₓ:H passivation layer stack, using the laser system outlined above with a wavelength of 532 nm, is around 350 mJ/cm² [178]. Most of the laser-induced defects are generated at the wafer surface [182], where the passivation layer is completely removed and the Al metal contacts are subsequently formed. These surface defects are not critical since the metal contacts itself induce a high recombination rate, which is only limited by thermal diffusion of charge carriers. However, it cannot be excluded that the passivation layer stack outside of the opened area, where the energy of the Gaussian-shaped laser pulse is not high enough to successfully remove the passivation layer, is also damaged. Moreover, even if no noticeable mechanical damage to the opened wafer surface is visible via optical microscopy or SEM analysis [175, 183], it cannot be excluded that damage was nevertheless introduced in the

![Figure 4.3: Schematic of the mechanism of indirect laser ablation of a dielectric layer on a Si substrate, using Gaussian-shaped laser pulses. After [178].](image)
Si substrate below the local contacts in the emitter, SCR and even in the base. RCs in the SCR, which lead to an increased $j_{02}$, could potentially explain the observed $FF$ loss. This issue is further discussed in section 4.3.

Figure 4.4 depicts the SEM image of the cross-section of a local contact manufactured by means of ps-laser ablation with a wavelength of 532 nm and a laser spot size $d_s \approx 20 \, \mu m$ on the flat rear side of a mono-facial n-PERT solar cell fabricated in this work (published in [60]). The co-diffused boron emitter is marked in red and the n-type base Si substrate is marked in yellow. The surface is covered by a 1 µm thick full-area Al layer fabricated by means of physical vapor deposition (PVD), which is marked in green. It is visible that the $\text{Al}_2\text{O}_3/\text{SiN}_x/\text{H}$ passivation layer stack is completely removed in the laser-treated area. Moreover, no noticeable mechanical laser damage is observed in the opened area and outside of the opened area. However, it is not possible to detect RCs in the Si substrate neither by optical microscopy nor by SEM analysis. Note the different dimensions of laser spot size $d_s \approx 20 \, \mu m$ and emitter depth $d_e \approx 0.5 \, \mu m$.

**Figure 4.4:** SEM image: Cross-section of a local contact on the rear side of a mono-facial n-PERT Si solar cell, manufactured by means of ps-laser ablation and subsequent full-area deposition of Al. The n-type base material is marked in yellow and the co-diffused boron emitter is marked in red. In the laser-treated area, marked in green and labeled with Laser, the $\text{Al}_2\text{O}_3/\text{SiN}_x/\text{H}$ passivation layer stack (marked in blue) is removed prior to the deposition of a 1 µm thick Al layer (marked in green). Published in [60].

### 4.2.2 Deposition of Al

In the previous subsection, LCO via ps-laser ablation and subsequent deposition of Al was introduced as an alternative approach for the formation of local contacts on boron emitters. As mentioned before, this method allows to use a smaller metallization fraction and avoids the formation of metal spikes, which occur for screen-printed Ag/Al metal pastes during contact firing (see section 3.4.3). The Al metallization can be deposited by full-area screen-printing of Al pastes [184-186] or by PVD of Al [42, 184]. Both technologies result in metal contacts with
high electrical quality in terms of a low specific contact resistance, which is much less dependent on the emitter depth $d_e$ and surface doping concentration $N_S$, compared to screen-printed Ag/Al metal contacts, as shown in section 4.3.2. This enables the use of shallow boron emitter profiles with low emitter saturation current density $j_{0e}$ and high implied $V_{oc}$. Therefore, mono-facial n-PERT solar cells with LCO have the potential for higher fill factor $FF$ and $V_{oc}$, compared to bifacial n-PERT solar cells with screen-printed Ag/Al contacts.

Full-area screen-printing of Al is an attractive low-cost and industrially feasible technology, which also removes potential laser damage beneath the local contact spots via alloying of the Al into the Si substrate. However, this method can suffer from the formation of voids during alloying, which compromises the series resistance $R_S$ of the solar cells and may also lead to the formation of shunts [187, 188]. Furthermore, the mechanisms of contact and void formation depend strongly on the set peak firing temperature $T_S$ during contact firing, paste composition of the commercial Al pastes and structure of LCO, such as the line width and segment length in the case of a dashed-lines structure (see section 4.2.3). Moreover, the metal contacts of front and rear side must be fired simultaneously, which may result in a compromise in the firing parameters due to potential differences in the requirements for each paste.

To avoid these potential drawbacks, PVD of Al is exclusively used in this work, although it is less cost-effective. By using Al PVD, contact formation on front and rear side can be decoupled. This enables contact firing of the screen-printed front side Ag contacts at a low $T_S$, which is mandatory to preserve the high passivation quality of the Al$_2$O$_3$/SiN$_x$:H passivation layer stack used in this work (see section 3.4.4). Furthermore, $\rho_e$ is almost independent of the applied LCO structure, as discussed in the following, and the formation of voids is essentially reduced, since no alloying of the Al and Si occurs. It is reported in [184] that PVD of Al can lead to higher $FF$ compared to full-area screen-printing of Al at comparable $V_{oc}$ and metallization fraction $f_c$. The high potential of LCO and subsequent PVD of Al has been demonstrated in [42] on mono-facial n-PERT solar cells with $\eta = 22.5\%$, using an advanced process sequence compared to the one used in this work. The use of a process sequence more comparable to the one used in this work, resulted in $\eta = 21.3\%$ for a common BBry-based emitter diffusion and an optimized fine-line metallization scheme, reported in [184].

In the following, the two different methods for PVD of Al applied in this work are discussed:

I. Sputtering of Al:

Sputtering of Al is conducted in an evacuated chamber via ions of the inert gas like Ar that knock-out atoms of the Al target. The released Al atoms move, driven by their kinetic energy, to the wafer which is mounted opposite the target and adhere to the wafer surface so that an Al layer grows on the wafer surface. Its thickness depends on the time of the sputtering process. In this work, the AJA ATC 2200 sputtering system from the manufacturer AJA International Inc. was used. However, this system has drawbacks, namely, that it is not possible to deposit more than one large-area (156.25 cm$^2$ or
243.36 cm²) wafer at the same time and low deposition rates of around 0.1 nm/s. This results in a process time of almost three hours per wafer. Nevertheless, sputtering of Al was used as reference process, as it is usually used for industrial in-line Al deposition.

II. Electron beam evaporation

Electron beam evaporation is another form of PVD. In this technique, the Al target is bombarded with an electron beam which is generated by a charged tungsten filament in an evacuated process chamber. The high energy of the bombarding electrons evaporates the Al target. The evaporated Al atoms adhere to the wafer surface and an Al layer grows. In this work, the Classic 570 L evaporation tool from the manufacturer Pfeiffer Vacuum was used. This tool enables a higher throughput, compared to the sputtering tool outlined above, since it is possible to process five large-area wafers simultaneously. For the applied voltage of 10 kV and electron beam current of 650 mA, the deposition rate is around 1.5 – 2 nm/s, resulting in a total process time of about 30 min, including vacuum and cooling steps. However, the high energetic electrons generate characteristic radiation, which damages the passivation layer stack. This damage can be completely cured by thermal annealing at 300 - 400°C for 10 - 30 minutes, as shown e.g. in [189].

The following experiment was conducted to clarify if the deposition of Al by electron beam evaporation leads to similar IV-characteristics of mono-facial n-PERT solar cells compared to the industrial feasible sputtering technology. Therefore, large-area (243.36 cm²) mono-facial n-PERT solar cells were fabricated, corresponding to the process sequence outlined in section 4.1. As base material, n-type Cz-Si wafers with a base resistivity of 6 Ωcm and a post-process thickness of ≈ 150 μm were used. The Al layer was deposited by electron beam evaporation and by sputtering for two solar cells each.

Figure 4.5 shows the open-circuit voltage \(V_{oc}\), fill factor \(FF\), series resistance \(R_S\) and conversion efficiency \(\eta\) of the mono-facial n-PERT solar cells, as a function of the annealing time at a set temperature of 400°C, which corresponds to a wafer temperature of about 350°C. It is observed that electron beam evaporation causes substantial damage to the \(\text{Al}_2\text{O}_3/\text{SiN}_x/\text{H}\) passivation layer stack, due to characteristic radiation, resulting in post-deposited \(V_{oc}\) values below 620 mV. It is evident, that this damage can be completely cured by thermal annealing at a set temperature of 400°C for 10 min in a nitrogen atmosphere. These observations correspond to former studies, e.g. presented by Schiele et al. in [189]. On the other hand, Al sputtering does not cause noticeable damage to the passivation layer stack. Moreover, the electrical contact between Al and emitter forms during annealing, which can be proven by the strong increase in \(FF\) due to a significant decrease of \(R_S\) for the first annealing step (compare Figure 4.5b/c).

The measurements indicate that an annealing time of 5 – 10 min at a set temperature of 400°C is optimal in terms of high \(V_{oc}\) and \(FF\), resulting in similar values for the mono-facial n-PERT solar cells with Al metallization by electron beam evaporation compared to sputtering. Since
the throughput is much higher for the electron beam evaporation tool, it is mainly used for the following experiments.

![Graphs showing comparison of sputtering and electron beam evaporation of Al for rear side metallization of mono-facial n-PERT solar cells.](image)

**Figure 4.5:** Comparison of sputtering and electron beam evaporation of Al for rear side metallization of mono-facial n-PERT solar cells: a) open-circuit voltage $V_{oc}$, b) fill factor $FF$, c) series resistance $R_s$ and d) energy conversion efficiency $\eta$. The damage induced by characteristic radiation during deposition of Al by electron beam evaporation can be cured by thermal annealing at 400°C for 10 min. The annealing is also required to form an appropriate contact between the 1 µm thick Al layer and the boron emitter. The lines are guides for the eye.

### 4.2.3 Influence of contact geometry

The following study is focused on the influence of the LCO geometry on the IV-characteristics of large-area (156.25 cm²) mono-facial n-PERT Si solar cells. Note that part of the following results is published in [60]. As base material, n-type Cz-Si substrates with $\rho_b \approx 6 - 8 \ \Omega cm$ and post-process thickness $W \approx 150 \ \mu m$ were used. The boron emitter was formed by co-diffusion from a boron-doped SiOx:B layer deposited using the PECVD tool *Plasmalab System 100* from the manufacturer *Oxford Instruments*. 4PP measurements revealed sheet resistivities $R_{sh}$ of the boron emitter of 80 Ω/sq. In this experiment sputtering was used for the deposition of the Al metallization. Figure 4.6 illustrates the three different LCO structures studied within this work.
The first structure is a closely packed point structure with a spot size $d_s = 20 \, \mu m$ (Figure 4.6a). For the second LCO structure, dashed lines with a segment length of 800 $\mu m$, distance between adjacent segments of 800 $\mu m$ and a line width of 50 $\mu m$ were fabricated by laser ablation, using a laser spot size of $d_s = 12 \, \mu m$ (Figure 4.6b). For the third structure, parallel lines with a line width of 20 $\mu m$, line length of 123 mm and a distance between adjacent lines of 600 $\mu m$ were produced by laser ablation, also using a laser spot size of $d_s = 12 \, \mu m$ (Figure 4.6c). The rear side metallization fraction $f_c$ was varied between 0.01% and 3% for the closely packed point structure.

![Schematic of the different LCO structures manufactured by ps-laser ablation](image)

**Figure 4.6:** Schematic of the different LCO structures manufactured by ps-laser ablation: a) closely packed point structure, b) dashed lines and c) parallel lines. In d) the overlap of the laser pulses required for the complete opening of the dashed lines and parallel lines structure, is illustrated. After [60].

Table 7 shows the IV-characteristics of the best mono-facial n-PERT solar cells for the different LCO geometries. In this experiment, the closely packed point structures with $f_c = 1\%$ and $f_c = 3\%$ result in the highest solar cell efficiency of 19.7%. If $f_c$ is reduced, the series resistance $R_S$ increases substantially, mainly due to the lateral resistance of the boron emitter but also due to an increased contact resistance, which lowers $FF$ of the solar cells. Especially for the lowest $f_c = 0.05\%$, $R_S$ exhibits extremely high values, also lowering the short circuit current density $j_{sc}$. $V_{oc}$ decreases only slightly for an increase in $f_c$ because of the slightly increasing recombination at the metal contacts.

Comparing the different LCO geometries shows that the use of the point structure is evidently beneficial in terms of a high $\eta$. Although the dashed lines structure leads to comparable values for $V_{oc}$, $FF$ is $2\%_{abs}$ lower (for comparable $f_c$). The major part of this difference in $FF$ (about 1.1%) can be explained by an increased $R_S$, however, the residual difference in $FF$ of almost
1% cannot be explained by $R_S$. This observation is even more pronounced for the solar cells with parallel line structure. In this case, $R_S$ is equal or even lower compared to $R_S$ of the solar cells with the point structure, but $FF$ is also $2\%_{\text{abs}}$ lower.

**Table 7:** IV-characteristics of the best mono-facial n-PERT solar cells with co-diffused PECVD boron emitter for different LCO geometries, measured with a flasher tool with an absorbing black back sheet. Part of the data published in [60].

<table>
<thead>
<tr>
<th>LCO geometry</th>
<th>$f_c$ [%]</th>
<th>$V_{oc}$ [mV]</th>
<th>$j_{sc}$ [mA/cm$^2$]</th>
<th>$FF$ [%]</th>
<th>$\eta$ [%]</th>
<th>$R_S$ [Ωcm$^2$]</th>
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</thead>
<tbody>
<tr>
<td>points</td>
<td>0.05</td>
<td>670</td>
<td>36.0</td>
<td>26.8</td>
<td>6.5</td>
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<td>667</td>
<td>38.0</td>
<td>75.5</td>
<td>19.1</td>
<td>0.92</td>
</tr>
<tr>
<td>points</td>
<td>1</td>
<td>666</td>
<td>38.3</td>
<td>77.4</td>
<td>19.7</td>
<td>0.56</td>
</tr>
<tr>
<td>points</td>
<td>3.3</td>
<td>662</td>
<td>38.4</td>
<td>77.7</td>
<td>19.7</td>
<td>0.60</td>
</tr>
<tr>
<td>dashed lines</td>
<td>3.2</td>
<td>662</td>
<td>38.4</td>
<td>75.7</td>
<td>19.3</td>
<td>0.82</td>
</tr>
<tr>
<td>lines</td>
<td>3.3</td>
<td>656</td>
<td>38.1</td>
<td>75.7</td>
<td>18.9</td>
<td>0.56</td>
</tr>
</tbody>
</table>

Based on this first indication, it is assumed that overlaying laser pulses, which are necessary for the dashed lines as well as the parallel lines structure (see Figure 4.6d), induce considerable damage beneath the local contact spots. The non-overlaying single laser pulses, used for the point structure, also induce damage below the local contact spots, though less distinct. It is believed that this laser-induced damage is deep enough to reach the SCR, causing an increased saturation current density of the second diode $j_{02}$. As shown in section 2.3.4, an increased $j_{02}$ leads to serious loss in $FF$, whereas $V_{oc}$ is hardly affected.

This assumption is supported by the fact that for the point structure with $f_c = 3.3\%$, the total number of non-overlaying laser pulses with a laser spot diameter of 20 µm amounts to almost 400,000. This is a large, but nevertheless more than one order of magnitude smaller number than the total number of overlaying laser pulses required for the dashed line structure (almost 4,800,000 pulses), as well as for the line structure (almost 5,100,000 pulses). To prove this assumption and to get a deeper understanding of the relevant loss mechanisms due to the RCs which are potentially induced by local contact formation, further experiments were performed. The results of these experiments are presented and discussed in the next section. Based on the promising results mentioned above, the closely packed point structure was applied exclusively for the following experiments to achieve high efficiency of the mono-facial n-PERT solar cells.
4.3 Empirical model for laser-induced recombination centers

The idea of the following experiment is based on varying the boron emitter profiles and the metal fraction \( f_e \) on the boron emitter of mono-facial n-PERT solar cells and measuring the IV-characteristics of these solar cells. If the assumption that damage is generated underneath the local contact spots, possibly induced by the LCO and subsequent evaporation of Al, is true, noticeable differences in the IV-characteristics of these solar cells should be observed. The solar cells were fabricated corresponding to the process sequence introduced in section 4.1. Due to an improved process sequence and the utilization of the ICP-PECVD lab tool, the wafer size was extended to 6-inch, which enables the production of large-area (243.36 cm²) mono-facial n-PERT solar cells, satisfying industrial requirements. As base material, n-type mono-like Si wafers with \( \rho_b = 6 \Omega \text{cm} \) and a post-process thickness \( W \approx 200 \mu \text{m} \) and, as reference, n-type Cz-Si wafers with \( \rho_b = 9 \Omega \text{cm} \) and a post-process thickness \( W \approx 150 \mu \text{m} \) were used.

Mono-like Si wafers are manufactured by means of the low-cost Bridgman casting technique, which is applied for the crystal growth of multi-crystalline Si, using a mono-crystalline seed crystal at the bottom of the crucible [190-192]. This technique is also referred to as Seeding Cast Technique. The seed crystal defines the orientation of crystallization of the Si melt. Thus, the middle part of the ingot solidifies mono-crystalline with minimal dislocation incorporation, according to the orientation of the seed crystal, whereas multi-crystalline regions form in the boundary area. The crystallographic quality of the middle part of the mono-like ingots is comparable to Cz-Si ingots, if the solidification process is well controlled [192]. This enables essential cost reduction of high-quality Si base material for the fabrication of solar cells with high energy conversion efficiencies.

Boron emitter and phosphorus FSF were formed by means of co-diffusion from a boron-doped SiO\(_2\):B layer in a POCl\(_3\) tube furnace in a two-step process, as described in section 2.5.1. The doping layer was deposited in a single chamber ICP-PECVD lab tool, using the reaction gases SiH\(_4\), CO\(_2\) and B\(_2\)H\(_6\):H\(_2\), as discussed in section 2.5.2. The deposition and diffusion parameters were adjusted to obtain different boron emitter profiles, while maintaining low \( j_{oc} < 50 \text{ fA/cm}^2 \) and high \( iV_{oc} \) values. Lifetime measurements on non-metalized samples at 1 sun illumination revealed \( iV_{oc} \) values of up to 680 mV for mono-like Si and up to 700 mV for Cz-Si reference wafers. This indicates the excellent electrical quality of the PECVD boron- emitters and the good crystallographic quality of the mono-like material, which is only slightly lower compared to the Cz-Si. The different deposition and co-diffusion parameters enable the investigation of the impact of the damage due to the local contact formation on the IV-characteristics of the solar cells for different boron emitter profiles. The sheet resistivities of the boron emitters, measured by a 4PP setup, are in the range of \( 80 - 190 \Omega \text{sq} \). The doping profiles of the boron emitters (shown in Figure 4.7), measured by an ECV setup (see section 2.5.4), revealed emitter depths in the range of 500 - 620 nm (where \( N_D \approx 10^{17} \text{ cm}^{-3} \)) with surface doping concentrations \( N_S \approx 0.55 - 2.6 \times 10^{19} \text{ cm}^{-3} \).
Based on the results of the previous experiment, a closely packed point structure with contact diameter $d_S = 20 \, \mu m$ was manufactured by ps-laser ablation on the boron emitter, targeting high energy conversion efficiency of the mono-facial n-PERT solar cells. To study the impact of the damage on the IV-characteristics of the solar cells, $f_c$ was varied in the range of 0.25 – 2.0%. The full-area deposition of Al was realized by electron beam evaporation to enable a high throughput, as outlined in section 4.2.2.

![Figure 4.7: Doping profiles of the boron emitters and phosphorus FSF of the mono-facial n-PERT solar cells manufactured by means of co-diffusion from a PECVD SiO$_x$:B doping layer (measured by ECV).](image)

In the following subsections, the 3D numerical simulation of mono-facial n-PERT solar cells, performed in this work to reproduce the measured IV-characteristics, is introduced. For the first time, an empirical model for the distribution of the RCs underneath the local contact spots, which was implemented in the 3D simulation to describe the influence of the damage by local contact formation on $FF$ and $V_{oc}$ of the solar cells, is presented. This model is validated by comparing the measured IV-characteristics with IV-characteristics extracted from simulation. Based on this model, the ideal $f_c$ in terms of high $\eta$ is simulated and a detailed free energy loss analysis is presented to obtain a deeper understanding of the relevant loss mechanisms.

### 4.3.1 Numerical simulation

The 3D numerical simulation of the mono-facial n-PERT solar cells with LCO was performed using Synopsys TCAD Sentaurus. If the front side geometry of the n-PERT solar cells is fixed because of the finger grid for front metallization (see section 3.3), an arbitrary rear side metal contact coverage $f_c$, which correlates to an arbitrary distance between the local contact spots, is challenging to implement due to the difficulty to determine a small unit cell that can map the whole solar cell exactly by symmetry. To solve this issue, a simplified homogeneous front side geometry with a completely transparent full-area metal contact was implemented, thus having
geometrical constraints only on the rear side of the solar cells. The closely packed point structure (see Figure 4.6a) was modelled by a square unit cell with two quarters of the local contact spots, located at opposite corners of the unit cell, as shown in Figure 4.8. This unit cell maps the whole solar cell exactly by symmetry, enabling the simulation for an arbitrary $f_c$.

**Figure 4.8**: Unit cell used for the 3D numerical simulation of the mono-facial RJ n-PERT solar cells using a simplified homogeneous front side geometry and two alternate quarters of the local contacts on the rear side. This unit cell maps the whole solar cell exactly by symmetry, enabling the simulation for an arbitrary rear side metal contact coverage.

The relationship between diameter of the contact spots $d_S$, coverage of the metallization $f_c$ and length/width of the unit cell $a$ is given by

$$f_c = \frac{\pi}{8} \cdot \left( \frac{d}{a} \right)^2. \quad (4.1)$$

According to equation (2.29) in section 2.3.3, the total series resistance $R_S$ of a mono-facial n-PERT solar cell consists of the series resistance of the front metallization $R_{S,f}$ (including the line resistance of the Ag metal busbars $R_{BB,f}$ and fingers $R_{l,f}$, as well as the contact resistance of the interface between Ag metal contacts and FSF $R_{c,f}$), lateral resistance of the FSF $R_{FSF}$, resistance of the base $R_b$, lateral resistance of the boron emitter $R_e$ and series resistance of the rear metallization $R_{S,b}$ (including the contact resistance of the interface between local Al contacts and emitter $R_{c,b}$ and lateral resistance of the full-area Al contact $R_{lat,b}$)

$$R_S = R_{BB,f} + R_{l,f} + R_{c,f} + R_{FSF} + R_b + R_e + R_{c,b} + R_{lat,b}. \quad (4.2)$$
The contact resistance of the rear metallization can be calculated by

\[ R_{b,c}(f_c) = \frac{\rho_c}{f_c}, \tag{4.3} \]

with the specific contact resistance \( \rho_c \) of the evaporated Al contacts on the boron emitter.

While the resistance of the base \( R_b \) and lateral resistance of the emitter \( R_e \) are described by the simulation \( (R_{sim}) \), the use of the simplified full-area contact geometry on the front side does not permit the description of the series resistance of front metallization \( R_{f,f} \) and lateral resistance due to charge carrier transport in the FSF \( R_{FSF} \). Therefore, \( R_{FSF}, R_{S,f} \) and \( R_{lat,b} \) (which can be neglected) must be manually implemented in the simulation as a constant \( \Delta R \), which is independent of \( f_c \). For the total series resistance follows

\[ R_S(f_c) = R_{sim}(f_c) + R_{c,b}(f_c) + \Delta R = R_{sim}(f_c) + \frac{\rho_c}{f_c} + \Delta R. \tag{4.4} \]

The parameters \( \rho_c \) and \( \Delta R \) were determined by fitting of the simulated series resistance to the measured series resistance for the whole range of \( f_c \), resulting in excellent agreement for each emitter profile. The quantification of the simulated and measured \( R_S \) at MPP was performed with the two-light intensity method (TLIM) [193] from IV-curves at different light intensities. Based on this, the 1 sun IV-curve was then corrected for \( R_S \), allowing the extraction of the pseudo FF.

Moreover, using the simplified front side geometry, the phosphorus FSF is completely covered by the full-area metal contact. Therefore, the surface recombination velocity of minority charge carriers at the phosphorus FSF \( S_{p,FSF} \) was adjusted to fit to the measured saturation current densities of the FSF \( j_{0,FSF} \) (accounting for the passivated and metallized areas of the FSF) for the different boron emitters, while using \( S_{n,FSF} = 2\cdot10^7 \) cm/s in all cases.

The measured base resistivity of 6 \( \Omega \)cm \( (N_D = 7.6\cdot10^{14} \) cm\(^{-3} \)), injection dependent post-process minority carrier lifetime of the n-type mono-like Si wafers and doping profiles of boron emitter and phosphorus FSF (see Figure 4.7) were implemented in the simulation. The surface recombination velocity of the boron emitter with \( \text{Al}_2\text{O}_3/\text{SiN}_x:H \) passivation layer stack was assumed to be \( S_{n,e} = S_{p,e}/10 = 25.8 \) cm/s according to the data presented in [128].

It has to be mentioned that, because of separate processing of each series with different boron emitter profile and technical issues during processing, noticeable differences were measured in post-process lifetime \( (525 - 700 \) \( \mu \)s) of the n-type mono-like Si wafers, the saturation current densities of the FSF \( (j_{0,FSF} = 80 - 110 \) fA/cm\(^2 \), emitter saturation current density \( j_{0,e} \), specific contact resistance of the rear side metallization \( (\rho_c = 2.5 - 11.5\cdot10^{-4} \) \( \Omega \)cm\(^2 \) and constant series resistance contribution \( (\Delta R = 0.47 - 0.61 \) \( \Omega \)cm\(^2 \)). This led to their individual implementation in the simulation for each series with different boron emitter profile.

The next step is to implement a model for the distribution of RCs, potentially induced by LCO and subsequent evaporation of Al, beneath the local contact spots in the simulation.
4.3.2 Modelling of the distribution of recombination centers underneath local contacts

The SRH formalism (introduced in section 2.3.2) is used to describe the recombination activity of the RCs, which are potentially induced by LCO. According to equations (2.14)-(2.16), single level RCs are fully characterized by the capture cross-sections of the RCs for electrons/holes $\sigma_n/\sigma_p$, the density of the RCs $N_i$ and their energy level $E_i$ within the bandgap. Since it is not trivial to experimentally obtain information on the physical nature of the RCs, the following assumptions have been made:

1. $E_i = 0$: This means that the energy level of the RCs is located in the middle of the bandgap, which maximizes the recombination rate.
2. $\sigma_n = \sigma_p$: The capture cross-sections of the RCs for electrons and holes are equal.
3. $N_i^* = \sigma_n N_i$: The effective density of the RCs $N_i^*$ is used as parameter instead of $\sigma_n$ and $N_i$, since the SRH recombination rate is proportional to $\sigma_n N_i$ (see equation (2.15)). The use of $N_i^*$ obviates the knowledge of the absolute values of $\sigma_n$ and $N_i$.

Furthermore, no information on the distribution of the RCs could be determined experimentally. However, it is highly probable that the distribution of the RCs underneath the local contact spots is continuously decreasing with depth. It is assumed that the density of the RCs correlates with the absorbed power density distribution of the ps-laser pulses (law of Lambert-Beer). Therefore, the implemented distribution of the RCs is an exponential decay function of depth with the relationship

$$N_i^*(z) = N_{i0}^* \cdot e^{-\left(\frac{z}{L_i}\right)}, \quad (4.5)$$

with the characteristic length of the distribution $L_i$, the effective surface density $N_{i0}^* = \sigma_n N_{i0}$ of the RCs and their surface density $N_{i0}$.

As discussed in section 4.2.1, the energy of the laser pulses must exceed the ablation energy threshold for sufficient removal of the dielectric passivation layer stack. Since Gaussian-shaped laser pulses are used in this work, RCs may also be generated outside of the area of sufficient removal of the passivation layer stack (in the following denoted as physical contact area), where the energy of the laser pulse is lower than the ablation energy threshold but potentially still high enough to generate RCs. Therefore, it is assumed that the lateral distribution of the RCs outside of the physical contact area is a Gaussian-shaped function with $\sigma = 5 \mu m$.

Figure 4.9 shows the defect density for the implemented distribution of the RCs. By using this model, the area of influence of the RCs is significantly larger than the physical contact area. Note that it is not possible to implement a circular-shaped distribution of RCs in the physical contact area into Synopsys TCAD Sentaurus. To get a good approximation, a polygon with the same area (centered at the contact) was implemented instead (see Figure 4.9).
**Figure 4.9:** Distribution of the recombination centers underneath the local contact spots, potentially induced by LCO. The implemented distribution is an exponential decay function in depth. Laterally, the distribution is a Gaussian-shaped function outside of the physical contact area. Note that the z-axis is scaled by a factor of 10 for better visibility.

The distribution of the RCs is chosen to be qualitatively consistent with the simulated local temperature distribution reported in [183], which is typically induced by a ps-laser pulse during ablation of the dielectric layers. The empirical model was adjusted to the experimental data to reproduce the IV-characteristics $pFF$, $FF$ and $V_{oc}$ simultaneously, because they are strongly affected by the RCs. The best conformity was achieved for $L_t = 0.2 \, \mu m$ and $N_{T0}^* = 25 \, cm^{-1}$. This distribution exhibits values for $N_T^*$ of $0.7 - 2 \, cm^{-1}$ in the SCR, which are comparable to values obtained for a full-area Al BSF with $N_T^* = 0.9 \, cm^{-1}$ reported in [194]. This gives first indication that the above-mentioned assumptions for the SRH recombination properties and distribution of the RCs are reasonable.

The cross-sectional cut through a local contact spot, illustrated in Figure 4.10, shows the impact of the RCs on the recombination rate $R$ beneath the local metal contacts at MPP (red line). For comparison, $R$ without the RCs is also shown (green line). Furthermore, the effective density of the RCs $N_T^*$ as a function of depth (blue line) as well as the doping profile of the PECVD boron emitter $N_D$ (black line) are illustrated. This data is published in [112].
Figure 4.10: Result of the 3D numerical simulation at MPP: Comparison of the recombination activity $R$ beneath the local contacts with recombination centers (red line) and without recombination centers (green line). Additionally, the distribution of the recombination centers $N_T^*$ (blue line) as well as the doping profile $N_D$ (black line) are illustrated. Note that $R$ and $N_D$ have the same scale. Published in [112].

Because Auger recombination is the limiting recombination mechanism in the highly-doped boron emitter, $R$ without RCs follows the boron emitter doping profile in the emitter region (compare green and black lines in Figure 4.10). Although the effective density of RCs exhibits its highest values in the emitter region (see blue curve in Figure 4.10), the recombination rate $R$ with RCs is only slightly higher than $R$ without RCs in this region (compare green and red line in Figure 4.10). This is primarily caused by the already high $R$ of the rear metal contacts but also by separation of the charge carriers at the pn-junction and the low photo-generation rate of electron-hole pairs in the rear emitter. Thus, the density of minority charge carriers is low in the emitter region, which results in a relatively low $R$ (according to equation (2.14)).

It is observed that $R$ has a peak with drastically high values in the SCR (two orders of magnitude higher than in the emitter region) when RCs are beneath the metal contacts. In contrary to that, $R$ exhibits a minimum without RCs in the SCR at the same depth. This proves the assumption that a large $j_{02}$ is induced by the RCs, which mainly impacts $FF$ of the mono-facial n-PERT solar cells. How strong this impact on $FF$ is, is clarified in the next subsection.

Moreover, the detrimental impact of the RCs exceeds the emitter and SCR region and induces a substantially increased $R$ also in the base below (up to three orders of magnitude higher than without RCs). The increased $R$ in the base, as well as the slightly increased $R$ in the emitter, result in an enhanced saturation current density of the first diode $j_{01}$, which mainly lowers $V_{oc}$ (see section 2.3.4). However, the main factor limiting the conversion efficiency of the mono-
facial n-PERT solar cells, which is correlated to the local contacts, can be attributed to the RCs in the SCR, as discussed in more detail in the following subsection.

To obtain a better understanding and visualization of the detrimental influence of RCs beneath the local contact spots, the 3D illustration of $R$ with/without RCs is shown in Figure 4.11a/b. The pn-junction is marked in red. The lateral distribution of $R$ demonstrates that the area of impact is significantly larger than the physical contact area, especially in the SCR and base.

![Figure 4.11](image)

**Figure 4.11**: Total recombination rate: a) with RCs underneath the local contact spots and b) without RCs beneath the local contacts. Without RCs, the recombination rate in the emitter is only caused by Auger recombination. By the implementation of RCs into the simulation, the recombination activity is substantially increased, especially in the SCR but also in the base underneath as well as outside of the physical contact area. Note that the z-axis is scaled by a factor of 10 for better visibility.

### 4.3.3 Comparison of simulation and experimental results

It was demonstrated in the previous subsection by numerical simulation that RCs underneath the local contact spots, caused by the LCO, induce a high $j_{02}$, which mainly lowers $FF$ of the solar cells. Although the above-mentioned assumptions for modeling of the distribution of the RCs are plausible, this model must be validated by comparison of simulated with measured IV-characteristics of the mono-facial n-PERT Si solar cells fabricated in this work. An important parameter is $pFF$, because it reflects solely the influence of recombination on $FF$ (the influence of $R_s$ is fully removed). Note that this is only valid for a negligible impact of the shunt resistance $R_P$ on the IV-characteristics. Since $R_P$ is clearly above 1 kΩcm$^2$ for all solar cells, its influence on $FF$ can be neglected.

Without RCs beneath the local contact spots, the simulated $pFF$ curves are almost flat (see Figure 4.12a). This corresponds to a recombination activity, which is almost independent of the contact coverage $f_c$. $R_s$ is a decreasing function of $f_c$, because it is dominated by the lateral

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resistance of the emitter and the contact resistance of the local contacts, which is proportional to $1/f_c$ (see equation (4.3)), being confirmed by TLIM measurements. Therefore, it is expected that $FF$ increases for an increasing $f_c$ (see dashed lines in Figure 4.12b). While this trend in $FF$ almost fits the measured data for low contact coverage ($f_c < 1\%$), the opposite trend is observed for higher $f_c$. This indicates that the expected increase in $FF$, due to a decreasing $R_S$ for increasing $f_c$, is overcompensated by an increasing recombination activity. This assumption is well confirmed by the decrease of the measured $pFF$ for increasing $f_c$, which can only be explained by a considerable increase of the recombination activity for an increase in $f_c$. This is consistent with the presence of RCs beneath the local contact spots, mainly increasing the recombination rate in the SCR and base below (see Figure 4.11), as discussed in the previous subsection.

The experimentally observed decrease in $pFF$ is well described by the empirical model for the distribution of the RCs (see Figure 4.12a) introduced above, enabling an accurate reproduction of the measured $pFF$.
of the measured $FF$ (see Figure 4.12b) for all different boron emitter profiles. It is observed, that the slope of the $pFF$ curve, as a function of $f_c$, is independent of the boron emitter profile. This observation leads to the assumption that the detrimental impact of enhanced recombination due to RCs underneath the local contact spots on $FF$, which can be mainly attributed to defects in the SCR and the thereby correlated enhanced $j_{02}$, is independent of the boron emitter profile. The deviating absolute values for the different boron emitter series can be explained by separate processing of each series and technical problems during processing. The lower level of $pFF$ for the shallower boron emitter profiles (160 $\Omega$/sq and 190 $\Omega$/sq) in comparison to the deeper boron emitter profile (80 $\Omega$/sq) can be explained by a thermal oxidation of 55 min instead of 5 min, which is required for the growth of the SiO$_2$ passivation layer on the FSF. This thick SiO$_2$ layer was removed and subsequently an additional thermal oxidation for 5 min was conducted. These multiple high temperature treatments can result in a degradation of the electrical quality of the bulk material and cause an increased saturation current density of the base $j_{0b}$. Moreover, the field effect passivation of the FSF (see section 2.6) is weakened due to the long oxidation time, which causes a marked decrease of the surface doping concentration.

The increasing recombination activity for an increasing $f_c$ not only compromises $FF$, but also $V_{oc}$ (see Figure 4.12c). The trend of $V_{oc}$ as a function of $f_c$ can also be accurately reproduced by the empirical model for the distribution of the RCs, which confirms its consistency. The slope of the $V_{oc}$ curve depends on the boron emitter profile. The impact of the RCs on $V_{oc}$ is stronger for the shallower emitter profiles. It is assumed that this phenomenon is caused by damaging of the passivation layer stack outside of the physical contact area. This observation corresponds to Figure 4.12, where an increased recombination rate is visible around the local contact spot. Due to a decreasing field-effect with decreasing emitter doping, it is believed that the detrimental influence of the RCs on $V_{oc}$ is more pronounced for the shallow emitter profiles.

The simulated conversion efficiency $\eta$ of the solar cells reproduces the measured values for the deep emitter profile (80 $\Omega$/sq) accurately, while $\eta$ is overestimated in the simulation for the shallower emitter profiles (160 $\Omega$/sq and 190 $\Omega$/sq). This can be explained by an overestimation in $j_{sc}$. Since $j_{sc}$ is completely unaffected by the RCs in experiment and simulation (data not shown) and $R_s < 1$ $\Omega$cm$^2$ for all solar cells, the observed overestimation can only be explained by optical effects such as shading and ARC issues. However, despite the uncertainties in $j_{sc}$, the trend in $\eta$ is reproduced.

In Table 8, the results of the best large area (243.36 cm$^2$) n-PERT RJ solar cells with local contacts are shown. The values were independently certified by ISE CalLab. The best mono-like n-PERT solar cell ($R_{oh} = 80$ $\Omega$/sq, $f_c = 0.5\%$) exhibits an efficiency of 20.2\% with a $V_{oc}$ of 661 mV. The best Cz reference n-PERT solar cell ($R_{oh} = 130$ $\Omega$/sq, $f_c = 1\%$) has an efficiency of 20.5\% with a $V_{oc}$ of 672 mV. The high $V_{oc}$ values demonstrate the excellent electrical quality of the boron emitter via co-diffusion from PECVD doping layers. Moreover, the comparable results of the low-cost mono-like n-PERT cells show the high potential of the alternative Si
material by means of *Seeding Cast Technique*. As far as the author knows, this is the highest efficiency presented on large area (243.36 cm²) mono-like Si material so far.

Table 8: Results of the best large area (243.36 cm²) n-PERT RJ solar cells with LCO [112].

<table>
<thead>
<tr>
<th>Material</th>
<th>$V_{oc}$ [mV]</th>
<th>$j_{sc}$ [mA/cm²]</th>
<th>FF [%]</th>
<th>$\eta$ [%]</th>
<th>pFF [%]</th>
<th>Rs [Ωcm²]</th>
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<td>Mono-like*</td>
<td>661.4</td>
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<td>78.57</td>
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<td>81.7</td>
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<td>Cz ref*</td>
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<td>39.14</td>
<td>77.95</td>
<td>20.51</td>
<td>81.2</td>
<td>0.77</td>
</tr>
</tbody>
</table>

*Independently certified by ISE CalLab.

4.3.4 Simulation of optimal contact geometry and free energy loss analysis

Based on the measured IV-characteristics of the mono-facial n-PERT solar cells and the results of the numerical simulation presented in the previous subsection, the optimal contact coverage $f_c$ for the closely packed point structure with RCs underneath the local contacts should be in the range of 0.5 - 1% for a contact diameter of $d_S = 20 \, \mu m$, while higher $\eta$ can be reached for higher $f_c$ without RCs. Figure 4.13 illustrates the correlation of $pFF$, $FF$ and $R_s$ and Figure 4.14 of $V_{oc}$ and $\eta$, extracted from numerical simulation of mono-facial n-PERT solar cells with/without RCs underneath the local contact spots, for a wider range of $f_c$. Moreover, the contact diameter $d_S$ is varied between 10 and 40 µm for the 80 Ω/sq PECVD boron emitter (left side) and the 190 Ω/sq boron emitter (right side).

It is observed that the detrimental influence of the RCs on the IV-characteristics of the solar cells is even more pronounced for smaller contact spots. For a smaller contact diameter $d_S$ and fixed $f_c$, the total count of local contacts is strongly increased, since the area of the contact spots scales with the square of the contact radius. This means that a decrease in $d_S$ from 20 to 10 µm requires a four times higher total count of local contacts to reach the same $f_c$. Thus, the total area of impact of the RCs, which was proven to be significantly larger than the physical contact area (compare Figure 4.11), considerably increases when $d_S$ is lowered. This also explains why $FF$ of the mono-facial n-PERT solar cells with parallel line and dashed line structure is lower compared to the closely packed point like structure (see section 4.2.3). On the other hand, an increase in $d_S$ from 20 to 30/40 µm lowers the impact of the RCs on the IV-characteristics. The ideal $f_c$ is thereby shifted towards higher values for an increasing $d_S$. The difference in $FF$ for low $f_c$ can be explained by the increasing emitter resistance for increasing $d_S$ at a fixed $f_c$.

It has to be mentioned that this behavior is only valid for the distribution of RCs used in this work, which extends the physical contact by 5 µm laterally, independent of $d_S$. This assumption has been made for simplicity, since no information on the lateral extension of the laser-induced damage as function of $d_S$ could be experimentally determined. Although, the distinct difference in the influence of the RCs on the IV-characteristics for different $d_S$ might be overestimated
especially for high $f_c$, it is plausible that this effect also occurs for the real distribution of RCs, when using Gaussian-shaped laser pulses.

Figure 4.13: $pFF$, $FF$ and $R_S$ of n-PERT solar cells using mono-like-Si, extracted from the 3D numerical simulations with/without recombination centers underneath the local contact spots for a wide contact coverage range. Additionally, the contact diameter was varied in the range of 10 – 40 µm. The results for the 80 Ω/sq PECVD boron emitter are shown on the left side, whereas the results of the 190 Ω/sq PECVD boron emitter are depicted on the right side. The lines are a guide to the eyes.
Figure 4.14: $V_{oc}$ and $\eta$ of n-PERT solar cells using mono-like-Si, extracted from the 3D numerical simulations with/without recombination centers underneath the local contact spots for a wide contact coverage range. Additionally, the contact diameter was varied in the range of $10 - 40 \, \mu m$. The results for the 80 $\Omega$/sq PECVD boron emitter are shown on the left side, whereas the results of the 190 $\Omega$/sq PECVD boron emitter are depicted on the right side. The lines are a guide to the eyes.

Usually, recombination and resistive power losses are the relevant loss mechanisms in solar cells and are considered for the optimization of the solar cell architecture. Recombination losses are usually quantified by the saturation current density $j_0$, as outlined in section 2.3.2, while resistance related losses are commonly determined by $R_S$ and $R_P$, as discussed in section 2.3.3. For the optimization of the solar cells, in terms of maximizing the energy conversion efficiency, these two power loss mechanisms must be minimized separately. However, it is not trivial to estimate the relative influence of these two loss mechanisms on $\eta$ of the solar cells, since these two quantities have different units. This issue was solved for optimizing the front side metal grid of bifacial n-PERT solar cells in this work by computing the whole IV-curve corresponding to the two-diode model (see section 3.3.5). Unfortunately, it is not always possible to describe the solar cell by the two-diode model e.g. when optimizing an IBC solar cell structure.

To solve this issue and generally get a deeper understanding of the relevant loss mechanisms in solar cells, the free energy loss analysis (FELA), introduced by Brendel et al. in [195], can be
performed. Considering that every photo-generated free charge carrier initially has an energy equal to the bandgap energy $E_g$, the ultimate energy conversion efficiency of a solar cell $\eta_u$ can be defined by [196]

$$\eta_u = \frac{1}{p_{in}} \cdot \frac{1}{A} \int_V G \cdot E_g \, dV,$$

(4.6)

with the power density $p_{in} = 0.1$ W/cm$^2$ of the incoming radiation (according to the AM1.5g spectrum), illuminated solar cell area $A$ and generation rate $G$. The integration is conducted over the whole volume $V$ of the solar cell. $\eta_u$ includes all optical losses (see section 2.3.1), as well as losses due to thermalization. From the numerical simulation of the mono-facial n-PERT solar cells (shown in Figure 4.15) $\eta_u$ was calculated to be around 48.7%.

However, only the free energy $F$ of the charge carriers can be extracted from the solar cell. $F$ of an electron-hole pair equals the difference of the quasi-Fermi levels for electrons and holes

$$\Delta E_F = E_{Fn} - E_{FP},$$

(4.7)

The loss of free energy by recombination of electron-hole pairs depends on the region of the solar cell where the recombination occurs, since the difference in the quasi-Fermi levels varies locally. Therefore, it is not possible to reliably calculate the free energy loss rates by recombination via multiplication of the recombination currents with the voltage of the solar cell $V_{MPP}$. Instead, the recombination currents must be weighted with the local difference of the quasi-Fermi levels.

The area-normalized free energy $\dot{f}_{MPP} = F/A$ extracted per second from the solar cell at MPP, which equals the sum of the area-normalized free energy flux carried out by electrons and holes through the metal contacts, can be calculated by [195, 196]

$$\dot{f}_{MPP} = p_{MPP} = V_{MPP} \cdot j_{MPP},$$

(4.8)

equaling the maximum output power density of the solar cell. Dividing this quantity by $p_{in}$, enables the calculation of the energy conversion efficiency of the solar cell [196]

$$\eta = \frac{p_{MPP}}{p_{in}} = \eta_g - \eta_r - \eta_s - \eta_n - \eta_p - \eta_m,$$

(4.9)

with the photo-generation efficiency

$$\eta_g = \frac{1}{p_{in}} \cdot \frac{1}{A} \int_V G \cdot \Delta E_F \, dV,$$

(4.10)

the efficiency loss by recombination in the solar cell volume with local recombination rate $R$

$$\eta_r = \frac{1}{p_{in}} \cdot \frac{1}{A} \int_V R \cdot \Delta E_F \, dV,$$

(4.11)
the efficiency loss due to recombination at the surface (passivated and contacted regions) with the surface recombination rate $R_{\text{surf}}$, which corresponds to the minority charge carrier flow to the surrounding surface $S$ of the solar cell

$$\eta_s = \frac{1}{p_{\text{in}}} \frac{1}{A} \int_S R_{\text{surf}} \cdot \Delta E_F \ dS,$$

(4.12)

the efficiency loss during electron/hole transport (by definition referred to as electron/hole Joule loss) caused by collision of electrons/holes with lattice atoms with the current density $j_{n/p}$ and conductivity $\sigma_{n/p}$ of electrons/holes

$$\eta_{n/p} = \frac{1}{p_{\text{in}}} \frac{1}{A} \int_V |j_{n/p}|^2 / \sigma_{n/p} \ dV,$$

(4.13)

and the efficiency loss by series resistance due to the metallization, such as contact resistances and line/lateral resistances of front/rear contacts

$$\eta_m = \frac{1}{p_{\text{in}}} \cdot (R_{S,f} + R_{S,b}) \cdot j_{\text{MPP}}^2.$$

(4.14)

By subdividing the solar cell into regions of interest, such as emitter, FSF/BSF, base and surface (contacted and non-contacted), their above-described contributions to the total efficiency loss could be computed automatically during numerical simulation using a C++ library which was in-house developed by G. Micard.

To get a deeper understanding of the loss mechanisms in the mono-facial n-PERT solar cells, a FELA was performed out of the numerical simulation presented above. The main contributors to the total efficiency loss of the solar cells are shown in Figure 4.15, exemplary for the 190 $\Omega/$sq co-diffused PECVD boron emitter since the impact of the RCs is strongest for the shallow emitter profile. Figure 4.15a depicts the FELA for local contact spots with $d_S = 10 \ \mu$m with RCs underneath the contact spots and Figure 4.15b illustrates the case without RCs. For comparison, the case for local contacts with $d_S = 20 \ \mu$m with RCs is depicted in Figure 4.15c.

First, note that the efficiency calculated by the FELA ($\eta$ FELA in Figure 4.15) equals almost perfectly the conversion efficiency of the numerical simulation (black curve in Figure 4.15). This strongly validates the reliability of the FELA. The ultimate efficiency $\eta_u$, extracted from the numerical simulation of the mono-facial n-PERT solar cells, was calculated to be around 48.7% (see horizontal line in Figure 4.15). The difference between ultimate efficiency $\eta_u$ and photo-generated efficiency $\eta_g$ (see equation (4.10)) is referred to as entropy loss. This entropy loss, which can be attributed to a loss of free energy by thermal movement of the charge carriers, depends strongly on the injection level $\Delta n$ in the solar cell and decreases for an increase in $\Delta n$.

Thus, the observed increasing entropy loss for an increasing $f_c$ can be explained by enhanced recombination at the local contacts, which lowers $\Delta n$. Although this effect occurs also for the case without RCs, it is stronger in the presence of RCs (compare Figure 4.15a/b), since the RCs additionally lower $\Delta n$. 
Further contributors to the loss of free energy are the SRH recombination in emitter and base, recombination in the FSF, Joule loss of holes in emitter and base, as well as Joule loss at the contacts and FSF. Note that holes are minority charge carriers in the base and majority charge carriers in the emitter. Other loss mechanisms are negligible. The increasing Joule loss of the charge carriers at the metal contacts for a decreasing $f_c$ can be explained by an increase in the contact resistance, which is proportional to $1/f_c$ (see equation (4.3)) and independent if RCs are underneath the local contacts or not (compare Figure 4.15a/b). The increasing Joule loss of majority charge carriers in the boron emitter for decreasing $f_c$ can be explained by an increase of the lateral resistance of the emitter. Hence, this Joule loss is stronger for the contacts with $d_s = 20 \mu m$ compared to contacts with $d_s = 10 \mu m$, at the same $f_c$. While the free energy loss...
caused by recombination in the FSF is almost constant over the whole range of $f_c$ and independent of $d_S$ when RCs are present, it is slightly lower with RCs compared to the case without RCs.

As expected, the free energy loss due to SRH recombination in the emitter strongly increases for increasing $f_c$ when RCs are present underneath the local contact spots. This trend is more pronounced for the smaller contact spots (compare Figure 4.15a/c). Without RCs, the loss of free energy caused by recombination in the emitter can be neglected due to negligible Auger recombination (see Figure 4.15b and Figure 4.11). However, the substantial increase in Joule loss in the bulk for an increasing $f_c$ is even stronger than the free energy loss induced by enhanced recombination in the emitter. This increased Joule loss is a consequence of the increased recombination which is induced by the RCs, since the minority charge carriers must flow to the recombination sites. This correlation was also demonstrated by Micard et al. in [196].

In the bulk, the Joule loss of holes (minority charge carriers in the bulk) with/without RCs is depicted in Figure 4.16. It is observed that the Joule loss is especially high underneath the local contact spots in the presence of RCs, while the Joule loss of the minority charge carriers is only slightly increased beneath the local contact for the case without RCs.

Moreover, it is observed that the SRH recombination in the bulk is lower for the case with RCs, compared to the case without RCs. This can be explained by the fact that charge carriers that recombine at the RCs can no longer recombine in the base. The higher recombination at the RCs is almost completely compensated by a decreased recombination in the bulk. Indeed, the sum of these two recombination channels is only slightly enhanced with RCs, compared to the case without RCs (compare Figure 4.15a/b). Figure 4.17 depicts the recombination rate of holes in the unit cell. Without RCs, the recombination rate is almost homogeneous in the bulk. By including RCs in the simulation, the recombination rate of holes in the bulk is essentially decreased close to the rear side because of an increased recombination activity at the rear side, which is induced by the RCs. However, note that the essential difference in the IV-characteristics with/without RCs cannot be explained by the increased recombination rate of holes at the RCs, which is almost fully compensated by the decreased recombination rate of holes in the bulk, but mainly by the significantly increased Joule loss of holes in the base for increasing $f_c$.

The loss channels, caused by RCs underneath the local contact spots, are almost as limiting as the loss channels introduced using a non-optimized single screen-printing technology, leading to high $R_S$ values and relatively high shadowing. This issue can be solved by applying high narrow fingers by means of print-on-print technology, photolithography and/or plating, as discussed in section 3.3.
**Figure 4.16:** Joule loss of the minority charge carriers in the bulk: a) with RCs underneath the local contact spots and b) without RCs underneath the local contacts. Without RCs, the Joule loss is only slightly increased below the local metal contacts. By the implementation of RCs into the simulation, the Joule loss of the minority charge carriers is substantially increased, especially beneath the local contacts but also between the contacts. Note that the z-axis is scaled by a factor of 10 for better visibility.

**Figure 4.17:** Recombination rate: a) with RCs underneath the local contact spots and b) without RCs underneath the local contacts. Without RCs, the recombination rate is almost homogeneous in the bulk. When RCs are included in the simulation, the recombination rate in the bulk is essentially decreased close to the rear side, because of increased recombination activity at the rear side, which is induced by the RCs. Note that the z-axis is scaled by a factor of 10 for better visibility.
4.4 Summary

In this chapter, the mono-facial RJ n-PERT solar cell process was introduced. The key process is the local contact formation by means of ps-laser ablation of the rear side passivation layer stack on the boron emitter and subsequent deposition of Al. The ps-laser pulses remove the dielectric layers by indirect ablation. The use of ps-laser pulses with a wavelength of 532 nm, results in complete removal of the dielectric layer without visible mechanical damage in terms of melting of the wafer surface. However, thermal damage cannot be completely avoided.

Sputtering or electron beam evaporation of Al can be used to achieve full-area deposition of Al for the rear side metallization after laser ablation. The electron beam evaporation technique significantly damages the passivation layers by characteristic radiation. This damage can be completely cured by thermal annealing at 400°C for 5 - 10 min in a nitrogen ambient. Although thermal annealing is not necessary for the sputtering technology, since no visible damage is induced during deposition, it is required to form an appropriate electrical contact between the metal and the Si substrate in terms of a low contact resistance.

Moreover, it was demonstrated that the use of a closely packed point LCO geometry with a spot size of 20 µm is beneficial in terms of a high efficiency of the mono-facial n-PERT Si solar cells. In comparison, the application of a dashed lines and parallel lines structure results mainly in a significantly lower FF for the same contact coverage. This observation could be attributed to the generation of defects in the SCR, which leads to high $j_{02}$ values. This damage is notably higher for the dashed line and parallel line structure due to overlaying laser pulses, which are mandatory for complete removal of the passivation layer. Thus, the total amount of required laser pulses is one order of magnitude higher, compared to the closely packed point structure.

Large area (243.36 cm²) mono-facial n-PERT solar cells with a conversion efficiency of 20.2% were presented in this chapter, utilizing potentially low-cost n-type mono-like Si. The Cz-Si reference solar cells exhibit an efficiency of 20.5%. The results demonstrate the excellent electrical quality of the boron emitters, fabricated by the industrially feasible co-diffusion via ICP-PECVD-deposited SiOₓ:B doping layers. The origin of the experimentally observed decrease in $pFF$, $FF$ and $V_{oc}$ for increasing contact coverage has been explained by recombination centers underneath the local contact spots, potentially induced by ps-laser ablation of the dielectric layer stack on the boron emitter. For the first time, an empirical model for their distribution was presented, which describes their influence on the IV-characteristics of the mono-facial n-PERT Si solar cells. With this empirical model, the absolute values and the trends of $pFF$, $FF$ and $V_{oc}$ were accurately reproduced by 3D numerical simulations. It was stated that the main loss of $FF$ can be attributed to defects which are located in the SCR, inducing a high $j_{02}$. It was shown that this $FF$ loss, due to an enhanced $j_{02}$, is independent of the boron emitter profile. Moreover, it was demonstrated that the contact diameter of the local contact spots has a significant influence on the IV-characteristics, when assuming traps underneath the local metal contacts. Small local metal contacts should be avoided due to the strongly increased total count of required laser.
pulses compared to larger contact spots for the same metal coverage, since the area of influence of the RCs extends the physical contact area substantially.

A detailed free energy loss analysis was presented, which clarifies the relevant loss mechanism in the mono-facial n-PERT solar cells. It was demonstrated that the essential difference in the IV-characteristics with/without RCs cannot be explained by the increased recombination rate of holes at the RCs, which is almost compensated by the decreased recombination rate of holes in the bulk, but mainly by the significantly increased Joule loss in the base for increasing $f_c$, which is a consequence of increased recombination at the RCs.

Furthermore, it was proven that the defects in the base and damage of the passivation layer outside of the local contact spots result in an increased $j_{01}$, which mainly lowers $V_{oc}$. Contrary to the influence of defects in the SCR on $FF$, the influence of these defects on $V_{oc}$ is dependent on the boron emitter profile. The loss of $V_{oc}$ for increasing contact coverage is stronger for a shallow emitter profile. Due to the increased detrimental impact of the recombination centers on the IV-characteristics of the solar cells, for increasing contact coverage, only a narrow range of contact coverage values between 0.5 - 1.0% is ideal for a spot size of 20 $\mu$m, in terms of a high conversion efficiency of the solar cells. This optimal contact coverage is almost independent of the boron emitter profile, since the influence of the recombination centers in the SCR dominates, confirming an independent behavior of the doping profile. Higher efficiencies and a wider range for the ideal contact coverage are expected from the simulation without the compromising influence of the recombination centers. While these defects can be partly attributed to the laser process used within this work, which is not yet fully optimized, they might not be completely avoidable, even with a more optimized process. As far as the author knows, no such study has yet been performed. The higher conversion efficiencies of mono-facial n-PERT solar cells reported in [42, 184] on a similar process sequence can be explained by a more advanced front metallization scheme. Indeed, the power loss which can be attributed to the front side metallization is the biggest contributor to the total power loss of the n-PERT Si solar cells presented in this work.
5 Summary

In this thesis, metallization-induced power losses of bifacial n-PERT Si solar cells were studied in detail since screen-printing metallization still limits the energy conversion efficiency $\eta$ of these solar cells. Thereby the focus was set on formation of Ag/Al metal spikes during contact firing and their detrimental influence on the IV-characteristics of bifacial n-PERT solar cells. For the first time, a detailed study on the impact of contact firing conditions on the formation of Ag/Al metal spikes was presented. It was demonstrated in this work that the area density (size and amount) of the Ag/Al metal spikes increases strongly for increasing set peak firing temperature $T_s$. Moreover, the presented microscopic SEM analysis revealed that the Ag/Al metal spikes are deep enough to easily penetrate the SCR of the solar cells even for low $T_s$, due to their average depth of at least 1 $\mu$m. This work shows that the detrimental influence of the Ag/Al metal spikes on the IV-characteristics of the solar cells depends strongly on the area density of the Ag/Al metal spikes. The increasing area density of the Ag/Al metal spikes for increasing $T_s$ agrees well with the decreasing specific contact resistance $\rho_c$ of screen-printed Ag/Al metal contacts on boron emitters for increasing $T_s$.

The impact of the Ag/Al metal spikes, in dependence on the boron emitter profile, was studied on bifacial n-PERT solar cell with different co-diffused boron emitters. It turned out in this study that the Ag/Al spike formation is independent of using an Al$_2$O$_3$/SiN$_x$:H passivation layer stack or an SiO$_2$/SiN$_x$:H passivation layer stack. As expected, the detrimental influence of the Ag/Al metal spikes on the saturation current density of the first diode $j_{01}$ of the solar cells for increasing $T_s$ is less distinct for solar cells with boron emitters with lower $R_{sh}$ (deeper emitter profiles with higher surface doping concentration) than for solar cells with boron emitters with higher $R_{sh}$, because of the better electrical shielding of the metal contacts. The slope of the $j_{01}$-$T_s$-curve is proportional to $R_{sh}$, therefore the detrimental influence of the Ag/Al spikes is less pronounced for boron emitters with lower $R_{sh}$. Since $j_{01}$ mainly limits the open-circuit voltage $V_{oc}$ of the solar cells, it is expected from these findings that $V_{oc}$ of solar cells with boron emitters with lower $R_{sh}$ decreases less than $V_{oc}$ of solar cells with higher $R_{sh}$ for increasing $T_s$. However, the opposite behavior was observed. This finding was explained by a substantial influence of the Ag/Al metal spikes, penetrating the SCR, on the saturation current density of the second diode $j_{02}$ of the solar cells. It turned out that the absolute values of $j_{02}$, as well as its increase for increasing $T_s$, is much more distinct for solar cells with boron emitters with lower $R_{sh}$ than for solar cells with boron emitter with higher $R_{sh}$. This explains the observed trend of $V_{oc}$, since such high $j_{02}$ values also affect $V_{oc}$ significantly. The $V_{oc}$ loss induced by $j_{02}$ was determined to be in the range of 5 – 30 mV, strongly depending on the boron emitter profile and $T_s$, being a noticeable part to the total $V_{oc}$ loss of the solar cells of 20 – 90 mV.

Although the series resistance $R_s$ of the solar cells is continuously decreasing for increasing $T_s$, which should result in a continuously increasing fill factor $FF$, it was observed that $FF$ is decreasing for high $T_s$. This $FF$ decrease is even stronger for the boron emitter profiles with lower
$R_{sh}$. This phenomenon can be explained by the strong impact of the Ag/Al metal spikes on $j_{02}$, mainly limiting $FF$ of the solar cells and thereby overcompensating the effect of the decreasing $R_s$. The Ag/Al spikes represent direct Schottky-contacts between the highly doped boron emitter and the metal. The width of the potential barrier is lower for a higher boron doping, resulting in a higher probability for charge carriers to tunnel through this potential barrier. Thus, parasitic leakage currents are induced. It is assumed that the detrimental leakage currents are proportional to the doping concentration, also explaining the lower $\rho_c$ observed for the boron emitters with lower $R_{sh}$. These findings agree well with existing analytical models for the calculation of $\rho_c$ published in [157, 163]. A potentially higher initial defect density in the boron emitters with lower $R_{sh}$ may additionally enhance parasitic leakage currents. Moreover, diffusion of paste components to the SCR and differences in the peak wafer temperature during contact firing for the different boron emitter profiles cannot be excluded.

The saturation current densities of the first and second diode in the two-diode model attributed to screen-printed Ag/Al metal contacts on boron emitters with $R_{sh} = 64 \, \Omega/sq$ and Al$_2$O$_3$/SiNx passivation layer stack (fired at $T_s = 800^\circ$C) were calculated to $j_{01c,Ag/Al} \approx 2,856 \, fA/cm^2$ and $j_{02c,Ag/Al} \approx 684 \, nA/cm^2$. These values are much higher than the ones determined for screen-printed Ag metal contacts on the phosphorus FSF with $R_{sh} = 40 \, \Omega/sq$ ($j_{01c,Ag} \approx 500 \, fA/cm^2$ and $j_{02c,Ag} \approx 20 \, nA/cm^2$), demonstrating the much higher recombination activity of screen-printed Ag/Al metal contacts. Based on these findings, the ideal Ag/Al metallization H-grid pattern was computed. The computation revealed the ideal parameters $w_F \approx 60 \, \mu m$ and $d_F = 1.6 - 2.0 \, mm$. These values are similar to the ones computed for the Ag metallization H-grid pattern on the front side. This indicates that the power losses caused by recombination at the Ag/Al metal contact are in the same range as the power losses due to shading of the Ag metal contacts. Using the optimized H-grid pattern, the industrial feasibility of sequential co-diffusion was confirmed on large area (156.25 cm$^2$) bifacial n-PERT Si solar cells with $\eta = 19.7\%$. This value is slightly lower than the values reported in [58, 173], which can mainly be explained by a non-advanced single screen-printing step metallization process used in this work.

To overcome the metallization-induced power losses attributed to the Ag/Al metal contacts, an alternative metallization scheme was additionally studied in this work. Contact formation of mono-facial n-PERT solar cells was established by local contact opening (LCO) by means of laser ablation. By this means the dielectric passivation layer stack is locally removed by ps-laser pulses with a wavelength of 532 nm. Subsequently, the metallization is conducted by full-area physical vapor deposition of Al layer with a thickness of around 1 $\mu m$. With this method $f_c$ can be reduced to a level of below 1%, in comparison to $f_c$ of screen-printed metal contacts in the range of 4 – 8%. Furthermore, the evaporated Al metallization does not require contact firing, thus obviating the formation of deep metal spikes. Therefore, this metallization scheme does not suffer from the issues related to Ag/Al metal contacts.

However, it turned out that LCO can also induce damage in the Si substrate beneath the local contacts, also reaching the SCR. This results also in an enhanced recombination activity, mainly
limiting $FF$ of the mono-facial n-PERT cells. The main loss of $FF$ can be attributed to defects located in the SCR, inducing a high $j_{02}$. It could be shown that this $FF$ loss, due to an enhanced $j_{02}$, is independent of the boron emitter profile, in contrary to the findings on the Ag/Al metal spikes where a strong dependence of the boron emitter profile was observed. For the first time, a model for the distribution of these laser-induced defects was presented in this work. This model allows to successfully simulate the detrimental influence of these defects on the IV-curves of the solar cells and accurately reproduces the measured IV-characteristics $pFF$, $FF$ and $V_{oc}$. It was demonstrated that the use of a closely packed point structure with a spot size $d_s = 20 \mu m$ is beneficial in terms of a high $\eta$ of the solar cells.

Moreover, it was proven in this work that the defects in the base and damage of the passivation layer outside of the local contact spots result in an enhanced $j_{01}$, mainly lowering $V_{oc}$. In contrary to the impact of the defects in the SCR on $FF$, the influence of these defects on $V_{oc}$ is dependent on the boron emitter profile. The $V_{oc}$ loss for increasing $f_c$ is stronger for boron emitters with higher $R_{sh}$. Due to the strongly increasing recombination activity of the defects for increasing $f_c$, only a narrow range of $f_c = 0.5 - 1.0\%$ shows up to be ideal for $d_s = 20 \mu m$. This ideal $f_c$ is almost independent of the boron emitter profile, since the influence of the defects in the SCR dominates. Higher $\eta$ and a wider range of $f_c$ are expected from the simulation without defects underneath the local contacts. Although these defects can be partly attributed to a not yet fully optimized laser process used in this work, they might not be completely avoidable, even with a more optimized process. As far as the author knows, no such study has yet been performed.

A detailed free energy loss analysis was presented to clarify the relevant loss mechanism in the mono-facial n-PERT solar cells. It was demonstrated that the considerable difference in the IV-characteristics with/without defects underneath the local contact spots cannot be explained by the increased recombination rate $R$ of holes at these defects, which is almost compensated by the decreased $R$ of holes in the bulk. The physical origin of this deviation is the essentially increased Joule loss of minority charge carriers (holes) in the base for increasing $f_c$, which is also a consequence of increased recombination at the defects.

The industrial feasibility of the alternative approach for boron emitter formation by means of sequential co-diffusion from doping layers manufactured via ICP-PECVD was again confirmed on large area (243.36 cm$^2$) mono-facial n-PERT solar cells with $\eta = 20.2\%$, using potentially low-cost n-type mono-like Si (Cz-Si reference: $\eta = 20.5\%$). The measured $V_{oc} = 670$ mV shows the high electrical quality of co-diffused boron emitters. The higher $\eta$ of mono-facial n-PERT solar cells reported in [42, 184] on a similar process sequence can mainly be explained by a more advanced front side metallization scheme. Indeed, the power losses attributed to the front side metallization are the biggest part of the total power loss of the n-PERT solar cells presented in this work. In conclusion, sequential co-diffusion shows up as a promising candidate for further reduction of production cost and is also applicable for advanced solar cell concepts like IBC.
6 Zusammenfassung

In dieser Arbeit wurden durch die Metallisierung induzierte Leistungsverluste von bifacialen n-PERT Si Solarzellen im Detail untersucht, da siebgedruckte Metallkontakte noch immer die Effizienz $\eta$ dieser Solarzellen limitieren. Dabei wurde der Fokus auf die Bildung von tiefen Ag/Al Metallkontakten während der Kontaktbildung in einem Hochtemperaturschritt und deren schädlichem Einfluss auf die IV-Kenngrößen der bifacialen n-PERT Solarzellen gelegt. In dieser Arbeit wurde erstmalig eine detaillierte Untersuchung des Einflusses der Parameter beim Kontaktfeuern auf die Bildung der Ag/Al Metallkontakte präsentiert. Es konnte gezeigt werden, dass die Flächendichte (Größe und Anzahl) der Ag/Al Metallkontakte mit steigender Feuer-temperatur $T_S$ stark zunimmt. Des Weiteren konnte durch die in dieser Arbeit vorgestellten REM Untersuchungen gezeigt werden, dass die Ag/Al Metallkontakte mit einer mittleren Tiefe von mindestens 1 µm bereits bei geringem $T_S$ tief genug sind, um die Raumladungszone der Solarzellen zu durchdringen. Der schädliche Einfluss der tiefen Ag/Al Metallkontakte auf die IV-Kenngrößen der Solarzellen hängt daher stark von deren Flächendichte ab. Die beobachtete Zunahme der Flächendichte der Ag/Al Metallkontakte erklärt den mit steigendem $T_S$ sinkenden spezifischen Kontaktwiderstand $\rho_c$ der siebgedruckten Ag/Al Metallkontakte auf Bor-Emittern.

Der Einfluss der Ag/Al Metallkontakte in Abhängigkeit von dem Profil der Bor-Emitter wurde anhand von bifacialen n-PERT Solarzellen mit unterschiedlichen kodiffundierten Bor-Emittern untersucht. Bei dieser Untersuchung wurde festgestellt, dass die Ag/Al Kontaktbildung unabhängig davon ist, ob ein Al$_2$O$_3$/SiN$_x$:H oder ein SiO$_2$/SiN$_x$:H Schichtsystem zur Passivierung verwendet wird. Der schädliche Einfluss der Ag/Al Metallkontakte auf die Sättigungsstromdichte der ersten Diode $j_{01}$ der Solarzellen mit steigendem $T_S$ ist, wie zu erwarten, weniger stark ausgeprägt für die Solarzellen mit Bor-Emittern mit geringem Schichtwiderstand $R_{sh}$ (tiefere Emitterprofile mit höherer Oberflächenkonzentration) als bei den Solarzellen mit Bor-Emittern mit hohem $R_{sh}$, aufgrund der besseren elektrischen Abschirmung der Metallkontakte. Die Steigung der $j_{01}$-$T_S$-Kurve ist proportional zu $R_{sh}$, wodurch der schädliche Einfluss der Ag/Al Metallkontakte für die Bor-Emitter mit geringerem $R_{sh}$ geringer ist. Da $j_{01}$ hauptsächlich die offene Klemmenspannung $V_{oc}$ der Solarzellen limitiert, wird auf Basis dieser Ergebnisse ein geringerer Abfall in $V_{oc}$ mit steigendem $T_S$ für die Solarzellen mit Bor-Emittern mit geringem $R_{sh}$ als bei den Solarzellen mit Bor-Emittern mit höherem $R_{sh}$ erwartet. Allerdings wurde das gegenteilige Verhalten beobachtet. Diese Beobachtung wurde durch einen erheblichen Einfluss der Ag/Al Metallkontakte, welche die Raumladungszone der Solarzellen durchdringen, auf die Sättigungsstromdichte der zweiten Diode $j_{02}$ erklärt. Es konnte in dieser Arbeit gezeigt werden, dass die absoluten Werte von $j_{02}$, sowie deren Zunahme mit steigendem $T_S$, für die Solarzellen mit Bor-Emittern mit geringerem $R_{sh}$ wesentlich ausgeprägter sind als für die Solarzellen mit Bor-Emittern mit höherem $R_{sh}$. Dadurch kann der beobachtete Verlauf von $V_{oc}$ erklärt werden, da hohe $j_{02}$ Werte ebenfalls $V_{oc}$ signifikant beeinträchtigen. Der durch $j_{02}$ verursachte Verlust in $V_{oc}$ liegt im Bereich von 5 – 30 mV und hängt stark vom Profil der Bor-Emitter und $T_S$ ab. Das ist ein beträchtlicher Anteil am totalen $V_{oc}$ Verlust der Solarzellen von 20 – 90 mV.

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Obwohl der Serienwiderstand $R_S$ der Solarzellen kontinuierlich mit steigendem $T_S$ abnimmt, woraus sich ein kontinuierlich steigender Füllfaktor $FF$ ergeben sollte, wurde eine Abnahme in $FF$ für hohe $T_S$ beobachtet. Diese Abnahme in $FF$ fällt stärker aus für die Bor-Emitter mit einem geringeren $R_{sh}$. Diese Beobachtung kann durch den starken Einfluss der Ag/Al Metallkontakte auf $j_{02}$, welche hauptsächlich den $FF$ der Solarzellen limitiert und dadurch den Effekt des abnehmenden $R_S$ überkompensiert, erklärt werden. Die Ag/Al Metallkontakte sind direkte Schottky-Kontakte zwischen dem hochdotierten Bor-Emitter und dem Metall. Die Breite der Potentialbarriere sinkt mit steigender Dotierkonzentration des Bor-Emitters, wodurch sich die Wahrscheinlichkeit des Tunnelns von Ladungsträgern durch die Potentialbarriere erhöht.

Dadurch werden parasitäre Verlustströme induziert. Es wird angenommen, dass die schädlichen Verlustströme proportional zur Dotierkonzentration des Bor-Emitters sind, wodurch sich auch der geringere $\rho_e$ der Bor-Emitter mit geringerem $R_{sh}$ verstehen lässt. Diese Ergebnisse stimmen sehr gut mit existierenden analytischen Modellen zur Berechnung von $\rho_e$ überein [157, 163].

Eine initial erhöhte Defektdichte in den Bor-Emittern mit geringerem $R_{sh}$ kann die parasitären Verlustströme zusätzlich begünstigen. Zusätzlich können eine Diffusion von Bestandteilen aus der Siebdruckpaste zur Raumladungszone und Unterschiede in der Wafertemperatur für die unterschiedlichen Emitterprofile während der Kontaktfeuern nicht ausgeschlossen werden.

Die Sättigungsstromdichte der ersten und zweiten Diode im Zwei-Dioden-Modell, welche den siebgedruckten Ag/Al Metallkontakten auf dem Bor-Emmitter mit $R_{sh} = 64 \, \Omega/$sq und dem Al$_2$O$_3$/SiN$_x$ Schichtstapel zur Passivierung (gefeuert bei $T_S = 800^\circ$C) zugeordnet werden kann, wurde zu $j_{01c,Ag/Al} \approx 2.856 \, fA/cm^2$ und $j_{02c,Ag/Al} \approx 684 \, nA/cm^2$ berechnet. Diese Werte sind viel höher als die Werte, welche für siebgedruckte Ag Metallkontakte auf einem Phosphor FSF mit $R_{sh} = 40 \, \Omega/$sq bestimmt wurden ($j_{01c,Ag} \approx 500 \, fA/cm^2$ und $j_{02c,Ag} \approx 20 \, nA/cm^2$). Das zeigt die deutlich höhere Rekombinationsaktivität der siebgedruckten Ag/Al Metallkontakte. Basierend auf diesen Resultaten wurde die ideale Grid-Struktur der Ag/Al Metallisierung berechnet. Die Berechnung führte zu den optimalen Parametern $w_F \approx 60 \, \mu m$ und $d_F = 1.6 - 2.0 \, mm$. Diese Werte sind mit den Werten vergleichbar, welche für die ideale Struktur der Ag Metallisierung auf der Vorderseite der Solarzellen berechnet wurde. Das deutet darauf hin, dass die Leistungsverluste durch Rekombination an den Ag/Al Metallkontakten in der selben Größenordnung sind wie die Leistungsverluste aufgrund von Abschattung durch die Ag Metallisierung. Unter der Verwendung der idealen Grid-Struktur konnte die industrielle Umsetzbarkeit der sequentiellen Kodiffusion auf großflächigen (156,25 cm$^2$) bifacialen n-PERT Si Solarzellen mit $\eta = 19,7\%$ bestätigt werden. Dieser Wert ist etwas geringer als der in [58, 173] veröffentlichte Wert. Das kann hauptsächlich durch den in dieser Arbeit verwendeten einfachen Siebdruckprozess zur Metallisierung der Solarzellen erklärt werden.

Um die durch die Ag/Al Metallisierung verursachten Leistungsverluste zu vermeiden, wurde in dieser Arbeit eine alternative Methode zur Kontaktierung von Bor-Emittern untersucht. Die Kontaktbildung von mono-facialen n-PERT wurde durch lokale Kontaktöffnung (LCO) mittels Laserablation realisiert. Bei dieser Methode wird die dielektrische Passivierschicht durch ps-

Es wurde allerdings festgestellt, dass die lokale Kontaktbildung mittels Laserablation ebenfalls zu einem Schaden im Si Substrat unterhalb der lokalen Kontakte führen kann, welcher die Raumladungszone erreichen kann. Das resultiert ebenfalls in einer erhöhten Rekombinationsaktivität, wodurch in erster Linie die FF der mono-facialen n-PERT Solarzellen limitiert wird. Der größte FF Verlust wird durch Defekte in der Raumladungszone verursacht, welche eine hohe $j_{02}$ induzieren. Es konnte gezeigt werden, dass dieser durch eine erhöhte $j_{02}$ verursachte FF Verlust unabhängig vom Profil der Bor-Emitter ist. Das steht im Gegensatz zu den Ergebnissen der Untersuchung der siebgedruckten Ag/Al Metallkontakte, bei denen eine ausgeprägte Abhängigkeit vom Profil der Bor-Emitter beobachtet wurde. In dieser Arbeit wurde erstmals ein Modell zur Beschreibung der Verteilung der durch die Laserablation generierten Defekte vorgestellt. Mit diesem Modell ließ sich der schädliche Einfluss der Defekte auf die IV-Kurven der Solarzellen erfolgreich simulieren und die gemessenen IV-Kenngrößen $pFF$, $FF$ and $V_{oc}$ erfolgreich exakt reproduzieren. Es wurde demonstriert, dass der Einsatz einer dicht gepackten Struktur aus Punkten mit einer Punktgröße $d_5 = 20$ µm vorteilhaft für eine hohe Effizienz $\eta$ der Solarzellen ist.

Darüber hinaus konnte in dieser Arbeit gezeigt werden, dass die Defekte in der Basis und der Schaden der Passivierung außerhalb der der lokalen Kontakte in eine erhöhte $j_{01}$ resultieren, welche hauptsächlich $V_{oc}$ der Solarzellen verringert. Im Gegensatz zu dem Einfluss der Defekte in der Raumladungszone auf $FF$, ist der Einfluss der Defekte auf $V_{oc}$ vom Profil der Bor-Emitter abhängig. Der $V_{oc}$ Verlust mit steigendem $f_c$ ist starker ausgeprägt für die Bor-Emitter mit einem höheren $R_b$. Aufgrund der stark ansteigenden Rekombinationsaktivität der Defekte mit steigendem $f_c$, ist nur ein schmaler Bereich von $f_c = 0.5 - 1.0\%$ optimal für eine Punktgröße von $d_5 = 20$ µm. Dieser ideale Metallisierungsanteil ist nahezu unabhängig vom Profil des Bor-Emitters, da der Einfluss der Defekte in der Raumladungszone dominiert. Ein höherer $\eta$ und ein breiterer Bereich von $f_c$ wären ohne Defekte unterhalb der lokalen Kontakte möglich. Obwohl diese Defekte teilweise auf einen noch nicht vollständig optimierten Laserprozess zurückzuführen sind, kann nicht ausgeschlossen werden, dass Defekte auch bei einem optimalen Laserprozess generiert werden. Soweit es dem Autor bekannt ist liegt bisher keine solche Studie vor.

Eine detaillierte Verlustanalyse freier Energie wurde präsentiert, um die relevanten Verlustmechanismen der mono-facialen n-PERT Solarzellen zu bestimmen und zu quantifizieren. Es konnte gezeigt werden, dass der signifikante Unterschied der IV-Kenngrößen mit/ohne Defekte

Die industrielle Umsetzbarkeit des alternativen Ansatzes zur Bildung von Bor-Emittern durch sequentielle Kodiffusion aus dotierten Schichten, hergestellt mittels ICP-PECVD, wurde erneut auf großflächigen (243,36 cm$^2$) mono-facialen n-PERT Solarzellen mit $\eta = 20,2\%$ demonstriert, unter Verwendung von potentiell kosteneffizientem n-Typ mono-like Si (Cz-Si Referenz: $\eta = 20,5\%$). Das gemessene $V_{oc} = 670$ mV zeigt die hervorragende elektrische Qualität der kodiffundierten Bor-Emittern. Der höhere $\eta$ der mono-facialen n-PERT Solarzellen veröffentlicht in [42, 184] mit einer ähnlichen Prozessfolge kann größtenteils durch eine fortgeschrittene Methode zur Metallisierung der Vorderseite erklärt werden. In der Tat sind die Leistungsweluste durch die Vorderseiten-Metallisierung der größte Anteil am Gesamtleistungsverlust der n-PERT Solarzellen. In dieser Arbeit konnte die sequentielle Kodiffusion als ein vielversprechender Kandidat für die zukünftige Reduktion von Produktionskosten identifiziert werden. Die sequentielle Kodiffusion kann ebenfalls für die Entwicklung von fortgeschrittenenen Solarzellenkonzepten wie den IBC Solarzellen eingesetzt werden.
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List of Publications


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