



HAL
open science

Impact of compensation on boron and oxygen-related degradation of upgraded metallurgical-grade silicon solar cells

Maxime Forster, J. Degoulange, R. Einhaus, G. Galbiati, E. Rougieux, A. Cuevas, Erwann Fourmond, Mustapha Lemiti

► To cite this version:

Maxime Forster, J. Degoulange, R. Einhaus, G. Galbiati, E. Rougieux, et al.. Impact of compensation on boron and oxygen-related degradation of upgraded metallurgical-grade silicon solar cells. *Solar Energy Materials and Solar Cells*, 2014, 120, pp.390-395. hal-01490336

HAL Id: hal-01490336

<https://hal.science/hal-01490336>

Submitted on 12 Feb 2018

HAL is a multi-disciplinary open access archive for the deposit and dissemination of scientific research documents, whether they are published or not. The documents may come from teaching and research institutions in France or abroad, or from public or private research centers.

L'archive ouverte pluridisciplinaire **HAL**, est destinée au dépôt et à la diffusion de documents scientifiques de niveau recherche, publiés ou non, émanant des établissements d'enseignement et de recherche français ou étrangers, des laboratoires publics ou privés.

1
2
3
4 **Impact of compensation on the boron and oxygen-related**
5
6
7 **degradation of upgraded metallurgical-grade silicon solar cells**
8
9

10
11
12 Maxime Forster¹, Pierre Wagner^{1,2}, Julien Degoulange¹, Roland Einhaus¹, Giuseppe
13 Galbiati⁴, Fiacre Emile Rougieux³, Andrés Cuevas³, Erwann Fourmond² and
14
15 Mustapha Lemiti²
16
17

- 18
19
20 1. APOLLON SOLAR, 66 cours Charlemagne, 69002 Lyon, France
21
22 2. INSA de LYON, INL, 7 av. J. Capelle, 69621 Villeurbanne Cedex, France
23
24 3. ISC Konstanz
25
26
27 4. Research School of Engineering, The Australian National University, Canberra
28
29 ACT 0200, Australia
30
31
32
33

34 **Abstract:**
35

36
37 This paper deals with the impact of dopant compensation on the degradation of
38
39 carrier lifetime and solar cells performance due to the boron-oxygen defect. The
40
41 boron-oxygen defect density evaluated by lifetime measurements before and after
42
43 degradation is systematically found proportional to the total boron concentration,
44
45 showing that compensation cannot reduce light-induced degradation. This result is
46
47 confirmed by a comparison of upgraded-metallurgical grade silicon solar cells having
48
49 identical boron, oxygen and carbon but different compensation levels and in which
50
51 the degradation is found more severe when the compensation is stronger.
52
53
54
55
56
57

58 **Keywords:**
59
60
61
62
63
64
65

1
2
3
4
5
6
7
8
9
10
11
12
13
14
15
16
17
18
19
20
21
22
23
24
25
26
27
28
29
30
31
32
33
34
35
36
37
38
39
40
41
42
43
44
45
46
47
48
49
50
51
52
53
54
55
56
57
58
59
60
61
62
63
64
65

Czochralski growth, dopant compensation, boron, phosphorus, gallium, oxygen, light-induced degradation, solar cells.

1 Introduction

Simultaneous presence of boron (B) and oxygen (O) in crystalline silicon (Si) is known to result in the existence of a strong recombination center – the boron-oxygen (BO) defect – which forms or is activated under excess carrier injection. In its active state, the BO defect dominates bulk recombination in B-doped cast or Czochralski (Cz)-grown Si, on which most of the photovoltaic industry is based. As a consequence, this defect currently represents one of the major wafer-related limitations of the performance of state-of-the-art industrial *p*-type Si solar cells [1].

Upgraded metallurgical-grade (UMG) Si solar cells are usually more affected by the BO defect, due to larger B concentration. Recent studies, however, found that the BO defect density was proportional in *p*-type Si to the net doping, rather than to the total B concentration, meaning that it could thus be reduced by compensation [2-5]. This finding was more recently contradicted by a study of the BO-related light-induced degradation of compensated Si solar cells, of which the severity was found to depend on the total B concentration [6]. With Ga co-doping offering an efficient tool for compensation engineering [7-9], it becomes important to clarify whether compensation can effectively be employed to mitigate the impact of the BO defect and thus improve the final performance of UMG-Si solar cells.

Another lately debated issue is the composition of the BO defect complex. Guided by observations that the BO defect density depends on the net doping – or majority carrier density – Voronkov and Falster [10] proposed a model in which the defect complex is made with *interstitial* boron (B_i). The dependence of the BO defect density was thereby explained by the proportionality of the solubility of positively charged B_i to the hole concentration, during the last phase of crystal cooling. Measurements in *n*-type compensated Si [11-13] and in Si co-doped with B and Ga [14] were nonetheless found

1
2
3
4 inconsistent with the involvement of B_i but rather suggested the participation of *substitutional* B. In
5
6 addition, recent calculations found that the concentration of B_i was too low to explain the observed
7
8 lifetime degradation and that the calculated characteristics of the B_iO_{2i} complex were inconsistent with
9
10 those measured by lifetime spectroscopy for the BO defect [15]. On the other hand, participation of
11
12 *substitutional* boron (B_s) in the BO defect complex remains very difficult to reconcile with the linearity
13
14 of the defect density to the net doping [16].
15
16
17
18

19 In the present work, we aim at verifying the effect of compensation on the BO recombination center
20
21 and clarifying the composition of the defect complex. In a first experiment, we study the BO-related
22
23 lifetime degradation in Si samples with a wide range of dopant concentrations and compensation levels.
24
25 We evaluate the oxygen-normalized effective BO defect density and find it to depend on the total boron
26
27 concentration, demonstrating the involvement of *substitutional* boron in the defect complex. In a
28
29 second experiment, we crystallize two different Cz ingots having identical B, O and C profiles but
30
31 different compensation levels. Then, we fabricate solar cells with wafers from both ingots and measure
32
33 their behavior under illumination. We find that the activation of the BO defect leads to stronger
34
35 performance degradation in solar cells made with heavily compensated Si than with lightly
36
37 compensated Si. This experiment unambiguously shows that compensation does not mitigate the
38
39 impact of the BO defect on solar cells but on the contrary worsens it.
40
41
42
43
44
45

46 **2 Experimental methods**

47 *2.1 Study of the lifetime degradation in compensated silicon wafers*

48
49 This study was carried out on *p*-type Cz-Si samples co-doped with Ga and B, B and P or B, P and
50
51 Ga. These samples originate from a batch of <100>-oriented Si ingots which were crystallized using
52
53 the Czochralski pulling technique. They were grown using electronic-grade (EG)-Si to which were
54
55
56
57
58
59
60
61
62
63
64
65

1
2
3
4 added various concentrations of B, P and Ga. In each sample, p_0 was deduced from Hall effect
5
6 measurement, assuming a Hall factor of $r_H = 0.75$ [17]. [B] was deduced from the sample's position in
7
8 the ingot, and from the initial concentration of B added into the Si melt, using Scheil's equation [18].
9
10 Together with uncompensated B-doped, P-doped and Ga-doped controls, these samples were
11
12 subsequently acid etched to remove saw damage, RCA cleaned, and underwent a phosphorus diffusion
13
14 gettering at 820°C to dissolve oxygen-related thermal donors and eliminate metallic impurities. This
15
16 ensures that there is no or very little impact of FeB and FeGa pairs during the degradation experiment.
17
18 Diffused regions were then removed by an additional acid etch and samples were RCA cleaned before
19
20 being coated on both sides by hydrogenated Si nitride ($\text{SiN}_x\text{:H}$) at 235 °C, using plasma-enhanced
21
22 chemical vapor deposition (PECVD).
23
24
25
26
27

28
29 After deposition, samples were annealed in the dark at 200 °C under air ambient during 30 min, in
30
31 order to annihilate the boron-oxygen defect. The effective lifetime was then measured using
32
33 inductively-coupled photo-conductance decay [19], immediately after annealing (τ_0) and after different
34
35 times t of illumination under a 10 mW/cm² halogen lamp at 25-27 °C.
36
37

38
39 A fit to Dannhauser and Krause's mobility sum measurements [20-22] is commonly used to convert
40
41 photo-conductance data into excess carrier density for injection-dependent lifetime measurements [19].
42
43 This fit is accurate in uncompensated Si [23] but overestimates the mobility sum in compensated Si [ref
44
45 Ziv]. To account for that lower mobility in our samples, we adjust the doping used as input in the fit to
46
47 Dannhauser and Krause's mobility so that it yields the same low-injection mobility sum as we
48
49 previously measured in our compensated samples (see [24]).
50
51
52

53
54 The effective lifetime degraded under illumination in all samples except for those which did not
55
56 contain B, in which it remained stable. This proves that the surface passivation quality of the $\text{SiN}_x\text{:H}$
57
58 layer did not change throughout the experiment. The observed degradation in samples containing B can
59
60
61
62
63
64
65

thus be attributed to a decrease of the bulk lifetime. After complete degradation, i.e. after saturation of the measured lifetime to its final value τ_{deg} , the effective BO defect density is calculated using the following expression:

$$N_t^* = \frac{1}{\tau_{\text{deg}}} - \frac{1}{\tau_0} \quad (1)$$

in which τ_{deg} and τ_0 are taken at a fixed injection level of $\Delta n = 0.1 \times p_0$. It is well known that the effective defect density after complete degradation $N_{t,\text{sat}}^*$ displays a quadratic dependence on $[\text{O}_i]$ [25]. This dependence is thought to be due to the participation of oxygen dimers which are themselves present in a concentration proportional to $[\text{O}_i]^2$ [26]. Since we intend to study the impact of dopants on the BO defect density, a more relevant parameter to focus on is:

$$N_{BC} = \frac{N_{t,\text{sat}}^*}{[\text{O}_i]^2} \quad (2)$$

This oxygen-normalized parameter reflects the influence of defect components (such as $[\text{B}]$ or p_0) not directly related to $[\text{O}_i]$. To evaluate N_{BC} in our samples, $\text{SiN}_x\text{:H}$ layers were etched off in hydrofluoric acid after the degradation experiment and $[\text{O}_i]$ was measured by Fourier transform infrared spectroscopy (FTIR).

2.2 Study of the degradation of compensated Si solar cells

a. Feedstock preparation

For this study, we used 30kg of PHOTOSIL UMG-Si [27-29]. The starting metallurgical-grade (MG)-Si was produced by FerroPEM, with careful selection of raw materials, i.e. silica and carbon, enabling to limit the initial impurity concentration. A first segregation treatment was applied during solidification of this MG-Si straight away after carbothermic reduction, decreasing the concentration of phosphorus and metallic impurities. The resulting material was then transferred to the PHOTOSIL

1
2
3
4 facilities where it was purified according to the standard PHOTOSIL process [27-29], involving a
5
6 plasma treatment to remove B and segregations to remove P and metallic impurities. After purification,
7
8 the Si feedstock contains 0.25 ppmw of B and 0.60 ppmw of P.
9

10
11 *b. Crystallization*
12
13

14 This lot of 30 kg of PHOTOSIL UMG-Si was divided into two batches of 15 kg, considered as
15
16 identical in terms of impurity concentration. These two batches were used to grow two different
17
18 <100>-oriented Si ingots of about 160 mm in diameter using the Czochralski crystallization technique.
19
20 Both crystals were pulled with an average rate of $v = 0.7$ mm/min, a crystal rotation rate of $w_{\text{cryst}} = 1.26$
21
22 rad.s^{-1} and an inverse rotation of the crucible with a rate of $w_{\text{cruc}} = 0.84$ rad.s^{-1} under an Ar pressure of
23
24 20 torr.
25
26
27

28 The first crystal, which we will refer to as Cz#U1, was grown without addition of any dopants aside
29
30 from those already contained in the UMG-Si. The second ingot, called Cz#U2, was grown using the
31
32 same crystallization process except for that we added 1.4 ppmw of P and 85 ppmw of Ga into the
33
34 crucible before melting of the Si. The addition of P enables to reduce the net doping while the addition
35
36 of Ga serves to maintain fully *p*-type ingot. Note that without the addition of Ga, adding P into the Si
37
38 melt would reduce the *p*-type fraction of the ingot and thus the ingot yield. After crystallization, both
39
40 ingots were shaped into 125×125 mm² pseudo-square bricks and the resistivity and carrier lifetime
41
42 profiles were measured respectively by Eddy current and $\mu\text{W-PCD}$ along the length of each brick.
43
44
45
46
47

48 *c. Solar cells fabrication and characterization*
49

50 Both ingots were then cut by a slurry wire saw into wafers with a thickness of about 200 μm . Hall
51
52 effect measurements were carried out on 2×2 cm² samples cut into wafers taken from different heights
53
54 along each ingot.
55
56
57
58
59
60
61
62
63
64
65

1
2
3
4 Industrial type screen printed solar cells with full area phosphorus front side diffusion and Al back
5
6 contact were fabricated at ISC Konstanz with 40 wafers of each one of the two ingots, evenly
7
8 distributed along their length. Uncompensated B-doped EG-Si control wafers were processed
9
10 simultaneously – 10 wafers with a resistivity of 1.5 $\Omega\cdot\text{cm}$ and 10 wafers with a resistivity of 3.0 $\Omega\cdot\text{cm}$.
11
12

13
14 The current-voltage (I-V) characteristics of these solar cells were then measured using a flash I-V
15
16 tester before degradation and after 60h under illumination at a temperature of 60°C.
17
18

19 20 **3 Results and discussion**

21 22 *3.1 BO defect density in compensated Si wafers*

23
24 To observe separately the potential influences of [B] or p_0 on N_{BC} , we selected among our samples
25
26 groups with equal p_0 and other groups with equal [B]. To study the influence of [B], we found two
27
28 groups of samples with iso- p_0 : one with 7 samples having $p_0 = (2.0 \pm 0.3) \times 10^{16} \text{ cm}^{-3}$ and one with 3
29
30 samples with $p_0 = (5.7 \pm 0.4) \times 10^{16} \text{ cm}^{-3}$. Plotting N_{BC} against [B] in these two groups (Figure 1-(a))
31
32 reveals a clear correlation between these two quantities. To study the influence of p_0 , we found two
33
34 groups of 4 samples, the first one with [B] = $(1.7 \pm 0.06) \times 10^{17} \text{ cm}^{-3}$ and the second one with [B] =
35
36 $(3.6 \pm 0.6) \times 10^{16} \text{ cm}^{-3}$. In Figure 1-(b), we see an increase of N_{BC} with p_0 in samples with [B] = $1.7 \pm$
37
38 $0.06 \times 10^{17} \text{ cm}^{-3}$ and no clear trend in samples with [B] = $(3.6 \pm 0.6) \times 10^{17} \text{ cm}^{-3}$.
39
40
41
42
43

44 The clear correlation between N_{BC} and [B] unambiguously indicates the participation of
45
46 *substitutional* boron B_s in the BO defect complex, since there is no other means through which [B] can
47
48 influence N_{BC} . We have proposed in [14], as an alternative explanation to the involvement of B_s in the
49
50 complex, that $[\text{B}_i]$ or $[\text{O}_{2i}]$ could be reduced due to the presence of Ga. This could have justified,
51
52 keeping B_iO_{2i} as the defect complex, the lower values for N_{BC} in Si co-doped with Ga and B than in
53
54 uncompensated B-doped Si with the same p_0 . Such a proposition can be discarded in the views of the
55
56
57
58
59
60
61
62
63
64
65

1
2
3
4 present results, since in samples co-doped with B, P and Ga and in which $p_0 < [B]$, N_{BC} is still
5
6 measured to be proportional to $[B]$. The only remaining plausible interpretation for the variation of N_{BC}
7
8 with $[B]$ is hence the participation of B_s in the BO complex.
9

10 11 3.2 Doping and lifetime profiles 12 13

14 As can be seen on Figure 2-(a), the net doping is significantly lower along most of ingot Cz#U2 than
15
16 along Cz#U1. This is due to compensation by larger concentration of P atoms in this ingot, while p -
17
18 type conductivity is maintained along the entire crystal length due to the presence of Ga. This lower net
19
20 doping leads to a higher resistivity, with an average of $2.5 \Omega \cdot \text{cm}$ along the first 70% of ingot Cz#U2
21
22 against $1.2 \Omega \cdot \text{cm}$ for ingot Cz#U1 (Figure 2-(b)).
23
24

25
26 Because both ingots Cz#U1 and Cz#U2 are grown with the same crystallization process, in the same
27
28 conditions and with the same Si feedstock, they have identical impurity profiles apart from P and Ga
29
30 which were added in Cz#U2. This means that two wafers taken from the same position in ingot Cz#U1
31
32 and Cz#U2 have the same B, O and C concentrations, but different p_0 . This will enable us to study in
33
34 the next section the impact of reducing p_0 by compensation on light-induced degradation of solar cells
35
36 without varying other parameters affecting the BO defect.
37
38

39
40 The carrier lifetime measured along both bricks is shown on Figure 3. Because of the smaller
41
42 majority carrier density in Cz#U2, the measured lifetime is about 3 times higher in this ingot than in
43
44 Cz#U1. This illustrates well the positive outcome of compensation on carrier lifetime before light-
45
46 induced degradation, which was previously reported and analyzed by several groups [30-32].
47
48

49 50 3.3 I-V characteristics 51

52 53 d. Before degradation 54

55 Measured I-V characteristics of fabricated solar cells are shown in Figure 4. As can be seen, solar
56
57 cells made with wafers from Cz#U2 have on average slightly better J_{sc} and slightly worse V_{oc} than
58
59
60
61

1
2
3
4 Cz#U1. The slightly better J_{sc} for Cz#U2 illustrates the positive impact of increasing compensation on
5
6 the minority carrier diffusion length. Because the lifetime enhancement due to compensation outweighs
7
8 the reduction in minority carrier mobility, the minority carrier diffusion length increases with
9
10 compensation [9], and so does J_{sc} . On the other hand, the lower V_{oc} for ingot Cz#U2 than for Cz#U1
11
12 illustrates that in the expression of V_{oc} (Equation 3), the reduction in p_0 due to compensation slightly
13
14 outweighs the increase in Δn resulting from the carrier lifetime improvement.
15
16
17
18

$$V_{oc} \approx \frac{kT}{q} \ln \left(\frac{\Delta n + (\Delta n + p_0)}{n_i^2} \right) \quad (3)$$

19
20
21
22
23 Overall, the differences in J_{sc} and V_{oc} between Cz#U1 and Cz#U2 cancel each other out and both
24
25 ingots yield similar solar cell conversion efficiencies η of respectively 17.94% and 17.92% on average.
26
27 This result shows that by using Ga co-doping, it is possible to tolerate a much higher P concentration (2
28
29 ppmw in Cz#U2 instead of 0.6 ppmw in Cz#U1) in the Si feedstock, without compromising neither the
30
31 ingot yield nor the solar cells performance. Such a finding is very positive because it strongly releases
32
33 the specifications on P concentration, of which the reduction represents an important cost element of
34
35 UMG-Si purification. Also note that the gap between the average efficiency of EG-Si and UMG-Si
36
37 solar cells is less than 0.3%, showing that UMG-Si is an appropriate material for making high-
38
39 efficiency solar cells.
40
41
42
43
44

45 *e. After degradation*

46
47
48 Figure 5 shows the efficiency of 6 pairs of solar cells taken from equivalent positions along ingots
49
50 Cz#U1 and Cz#U2 before and after light-induced degradation. Before light-induced degradation, solar
51
52 cells from both ingots have similar efficiency. After activation of the BO defect, the efficiency
53
54 degrades in both ingots but the degradation is significantly stronger in ingot Cz#U2, down to 16.4% on
55
56 average against 17% in Cz#U1. This confirms our previous finding that the BO defect density depends
57
58
59
60
61
62
63
64
65

1
2
3
4 on the total boron concentration rather than on the net doping. Both ingots Cz#U1 and CZ#U2 having
5
6 the same [B] profile, they also have the same BO defect density profile. Since the BO recombination
7
8 center is a deep level with a capture cross section for electrons which does not differ strongly from that
9
10 for holes [33] (capture cross section ratio of about 10 for the BO defect against around 1000 for
11
12 *interstitial* iron), its recombination activity does not vary much with the majority carrier density [34].
13
14 This means that carrier lifetime after degradation depends only on the BO defect density and should
15
16 therefore be the same in ingots Cz#U1 and Cz#U2. On the other hand, due to stronger compensation in
17
18 Cz#U2, the majority carrier density and the minority carrier mobility are lower in this ingot. This leads
19
20 to lower open-circuit voltage and short-circuit current in solar cells made with wafers from Cz#U2. On
21
22 the other hand, if the BO defect density was proportional to the net doping, carrier lifetime should be
23
24 twice longer in Cz#U2 than in Cz#U1 and both ingots should thus have similar efficiency.
25
26
27
28
29
30

31 32 **4 Conclusion** 33

34
35 In summary, we have shown through the analysis of the lifetime degradation in samples with a wide
36
37 variety of doping and compensation levels that the BO defect density is in general proportional to the
38
39 total boron concentration, rather than to the net doping. Such dependence implies the participation of
40
41 *substitutional* boron in the defect complex and means that light-induced degradation cannot be
42
43 mitigated by compensation. This finding is confirmed by a comparison between two UMG-Si Cz ingots
44
45 having identical B, O and C profiles but different compensation levels. Because both ingots have
46
47 identical carrier lifetime profiles after degradation the lower majority carrier density and minority
48
49 carrier mobility in the heavily compensated ingot lead to lower open-circuit voltage, short-circuit
50
51 current and hence conversion efficiency. This experiment unambiguously shows that whatever the
52
53 nature of the BO defect, its impact cannot be diminished by compensation.
54
55
56
57
58
59
60
61
62
63
64
65

1
2
3
4
5
6
7
8
9
10
11
12
13
14
15
16
17
18
19
20
21
22
23
24
25
26
27
28
29
30
31
32
33
34
35
36
37
38
39
40
41
42
43
44
45
46
47
48
49
50
51
52
53
54
55
56
57
58
59
60
61
62
63
64
65

1
2
3
4
5
6
7
8
9
10
11
12
13
14
15
16
17
18
19
20
21
22
23
24
25
26
27
28
29
30
31
32
33
34
35
36
37
38
39
40
41
42
43
44
45
46
47
48
49
50
51
52
53
54
55
56
57
58
59
60
61
62
63
64
65

Figure captions:

Figure 1: Normalized effective BO defect density plotted as a function of the total boron concentration (a) or as a function of the majority carrier density (b).

Figure 2: (a): Majority carrier density measured by Hall effect along UMG-Si ingots Cz#U1 and Cz#U2. (b): Resistivity measured by 4-point probe along UMG-Si ingots Cz#U1 and Cz#U2.

Figure 3: Carrier lifetime measured by μ W-PCD along ingots Cz#U1 and Cz#U2.

Figure 4: Open-circuit voltage (a), short-circuit current (b) and conversion efficiency of solar cells made with wafers from Cz#U1 and Cz#U2 and from EG-Si control wafers.

Figure 5: Measured solar cell conversion efficiency on solar cells along Cz#U1 and Cz#U2 before and after light-induced degradation.

- 1
2
3
4
5
6
7 [1] J. Schmidt, B. Lim, D. Walter, K. Bothe, S. Gatz, T. Dullweber, and P. P. Altermatt, "Impurity-
8 Related Limitations of Next-Generation Industrial Silicon Solar Cells," *Photovoltaics, IEEE*
9 *Journal of*, vol. PP, pp. 1-5, 2012.
- 10 [2] R. Kopecek, J. Arumughan, K. Peter, E. A. Good, J. Libal, M. Acciarri, and S. Binetti,
11 "Crystalline Si solar cells from compensated material: behaviour of light-induced degradation,"
12 in *Proceedings of the 25th European Photovoltaic Solar Energy Conference, Valencia, Spain*
13 *WIP, Munich*, 2008.
- 14 [3] D. Macdonald, F. Rougieux, A. Cuevas, B. Lim, J. Schmidt, M. D. Sabatino, and L. J. Geerligts,
15 "Light-induced boron-oxygen defect generation in compensated p-type Czochralski silicon,"
16 *Journal of Applied Physics*, vol. 105, p. 093704, 2009.
- 17 [4] B. Lim, F. Rougieux, D. Macdonald, K. Bothe, and J. Schmidt, "Generation and annihilation of
18 boron--oxygen-related recombination centers in compensated p- and n-type silicon," *Journal of*
19 *Applied Physics*, vol. 108, p. 103722, 2010.
- 20 [5] J. Geilker, W. Kwapil, and S. Rein, "Light-induced degradation in compensated p- and n-type
21 Czochralski silicon wafers," *Journal of Applied Physics*, vol. 109, p. 053718, 2011.
- 22 [6] S. Bernardini, Saynova, S. Binetti, and G. Coletti, "Light-Induced Degradation in compensated
23 mc-Si p-type solar cells," in *Proceedings 38th IEEE Photovoltaic Specialists Conference*,
24 Austin, 2012.
- 25 [7] J. Kraiem, R. Einhaus, and H. Lauvray, "Doping engineering as a method to increase the
26 performance of purified MG silicon during ingot crystallization," in *Proceedings of the 34th*
27 *IEEE Photovoltaic Specialists Conference*, Philadelphia, 2009, p. 1327.
- 28 [8] M. Forster, E. Fourmond, R. Einhaus, H. Lauvray, J. Kraiem, and M. Lemiti, "Ga co-doping in
29 Cz-grown silicon ingots to overcome limitations of B and P compensated silicon feedstock for
30 PV applications," *physica status solidi (c)*, vol. 8, pp. 678-681, 2011.
- 31 [9] M. Forster, B. Dehestru, A. Thomas, E. Fourmond, R. Einhaus, A. Cuevas, and M. Lemiti,
32 "Compensation engineering for uniform n-type silicon ingots," *Solar Energy Materials and*
33 *Solar Cells*, vol. 111, pp. 146-152, 2013.
- 34 [10] V. V. Voronkov and R. Falster, "Latent complexes of interstitial boron and oxygen dimers as a
35 reason for degradation of silicon-based solar cells," *Journal of Applied Physics*, vol. 107, p.
36 053509, 2010.
- 37 [11] F. E. Rougieux, B. Lim, J. Schmidt, M. Forster, D. Macdonald, and A. Cuevas, "Influence of
38 net doping, excess carrier density and annealing on the boron oxygen related defect density in
39 compensated n-article silicon," *Journal of Applied Physics*, vol. 110, p. 063708, 2011.
- 40 [12] F. E. Rougieux, M. Forster, D. Macdonald, A. Cuevas, B. Lim, and J. Schmidt, "Recombination
41 Activity and Impact of the Boron-Oxygen-Related Defect in Compensated N-Type Silicon,"
42 *Photovoltaics, IEEE Journal of*, vol. 1, pp. 54-58, 2011.
- 43 [13] V. V. Voronkov, R. Falster, K. Bothe, B. Lim, and J. Schmidt, "Lifetime-degrading boron-
44 oxygen centres in p-type and n-type compensated silicon," *Journal of Applied Physics*, vol. 110,
45 pp. 063515-063515, 2011.
- 46 [14] M. Forster, E. Fourmond, F. E. Rougieux, A. Cuevas, R. Gotoh, K. Fujiwara, S. Uda, and M.
47 Lemiti, "Boron-oxygen defect in Czochralski-silicon co-doped with gallium and boron,"
48 *Applied Physics Letters*, vol. 100, p. 042110, 2012.
- 49
50
51
52
53
54
55
56
57
58
59
60
61
62
63
64
65

- 1
2
3
4 [15] A. Carvalho, P. Santos, J. Coutinho, R. Jones, M. J. Rayson, and P. R. Briddon, "Light induced
5 degradation in B doped Cz-Si solar cells," *physica status solidi (a)*, vol. 209, pp. 1894-1897,
6 2012.
7
8 [16] D. Macdonald, A. Liu, A. Cuevas, B. Lim, and J. Schmidt, "The impact of dopant compensation
9 on the boron-oxygen defect in p- and n-type crystalline silicon," *physica status solidi (a)*, vol.
10 208, pp. 559-563, 2011.
11
12 [17] F. Szmulowicz, "Calculation of the mobility and the Hall factor for doped p-type silicon," *Phys.*
13 *Rev. B*, vol. 34, pp. 4031--4047, Sep 1986.
14
15 [18] E. Scheil, "Bemerkungen zur Schichtkristallbildung," *Z. metallk.*, vol. 34, pp. 70-70, 1942.
16
17 [19] R. A. Sinton and A. Cuevas, "Contactless determination of current--voltage characteristics and
18 minority-carrier lifetimes in semiconductors from quasi-steady-state photoconductance data,"
19 *Applied Physics Letters*, vol. 69, pp. 2510-2512, 1996.
20
21 [20] F. Dannhäuser, J. Krausse, and K. Mayer, "Zum einfluss von temperprozessen auf
22 widerstandsschwankungen in siliziumeinkristallen," *Solid-State Electronics*, vol. 15, pp. 1383-
23 1384, 1972.
24
25 [21] J. Krausse, "Die abhängigkeit der trägerbeweglichkeit in silizium von der konzentration der
26 freien ladungstrager - II," *Solid-State Electronics*, vol. 15, pp. 1377-1381, 1972.
27
28 [22] P. P. Altermatt, J. Schmidt, M. J. Kerr, G. Heiser, and A. G. Aberle, "Exciton-enhanced Auger
29 recombination in crystalline silicon under intermediate and high injection conditions," in
30 *Proceedings of the 16th European Photovoltaic Solar Energy Conference, Glasgow, Wales,*
31 2000.
32
33 [23] F. E. Rougieux, P. Zheng, M. Thiboust, J. Tan, N. E. Grant, D. H. Macdonald, and A. Cuevas,
34 "A Contactless Method for Determining the Carrier Mobility Sum in Silicon Wafers,"
35 *Photovoltaics, IEEE Journal of*, vol. 2, pp. 41-46, 2012.
36
37 [24] M. Forster, F. E. Rougieux, A. Cuevas, B. Dehstru, A. Thomas, E. Fourmond, and M. Lemiti,
38 "Incomplete Ionization and Carrier Mobility in Compensated p-Type and n-Type Silicon,"
39 *Photovoltaics, IEEE Journal of*, vol. 3, p. 108, 2013.
40
41 [25] J. Schmidt and K. Bothe, "Structure and transformation of the metastable boron- and oxygen-
42 related defect center in crystalline silicon," *Phys. Rev. B*, vol. 69, p. 024107, 2004.
43
44 [26] L. I. Murin, T. Hallberg, V. P. Markevich, and J. L. Lindström, "Experimental Evidence of the
45 Oxygen Dimer in Silicon," *Phys. Rev. Lett.*, vol. 80, pp. 93--96, Jan 1998.
46
47 [27] J. Kraiem, B. Drevet, F. Cocco, N. Enjalbert, S. Dubois, D. Camel, D. Grosset-Bourbange, D.
48 Pelletier, T. Margaria, and R. Einhaus, "High performance solar cells made from 100% UMG
49 silicon obtained via the Photosil process," in *Proceedings of the 35th IEEE Photovoltaic*
50 *Specialists Conference, Hawaiï*, 2010.
51
52 [28] D. Sarti and R. Einhaus, "Silicon feedstock for the multi-crystalline photovoltaic industry,"
53 *Solar Energy Materials and Solar Cells*, vol. 72, pp. 27-40, 2002.
54
55 [29] R. Einhaus, J. Kraiem, F. Cocco, Y. Caratini, D. Bernou, D. Sarti, G. Rey, R. Monna, C. Trassy,
56 J. Degoulange, Y. Delannoy, S. Martinuzzi, I. Périchaud, M. C. Record, and P. Rivat,
57 "PHOTOSIL - Simplified Production of Solar Silicon from Metallurgical Silicon," in
58 *Proceedings of the 21st European Photovoltaic Solar Energy Conference, Dresden, Germany,*
59 *WIP, Munich*, 2006.
60
61 [30] J. Libal, S. Novaglia, M. Acciarri, S. Binetti, R. Petres, J. Arumughan, R. Kopecek, and A.
62 Prokopenko, "Effect of compensation and of metallic impurities on the electrical properties of
63 Cz-grown solar grade silicon," *Journal of Applied Physics*, vol. 104, p. 104507, 2008.
64
65

1
2
3
4
5
6
7
8
9
10
11
12
13
14
15
16
17
18
19
20
21
22
23
24
25
26
27
28
29
30
31
32
33
34
35
36
37
38
39
40
41
42
43
44
45
46
47
48
49
50
51
52
53
54
55
56
57
58
59
60
61
62
63
64
65

[31] S. Dubois, N. Enjalbert, and J. P. Garandet, "Effects of the compensation level on the carrier lifetime of crystalline silicon," *Applied Physics Letters*, vol. 93, p. 032114, 2008.

[32] D. Macdonald and A. Cuevas, "Recombination in compensated crystalline silicon for solar cells," *Journal of Applied Physics*, vol. 109, pp. 043704-043704, 2011.

[33] S. Rein and S. W. Glunz, "Electronic properties of the metastable defect in boron-doped Czochralski silicon: Unambiguous determination by advanced lifetime spectroscopy," *Applied Physics Letters*, vol. 82, pp. 1054-1056, 2003.

[34] D. Macdonald and A. Cuevas, "Recombination in compensated crystalline silicon for solar cells," *Journal of Applied Physics*, vol. 109, p. 043704, 2011.

Figure 1 a

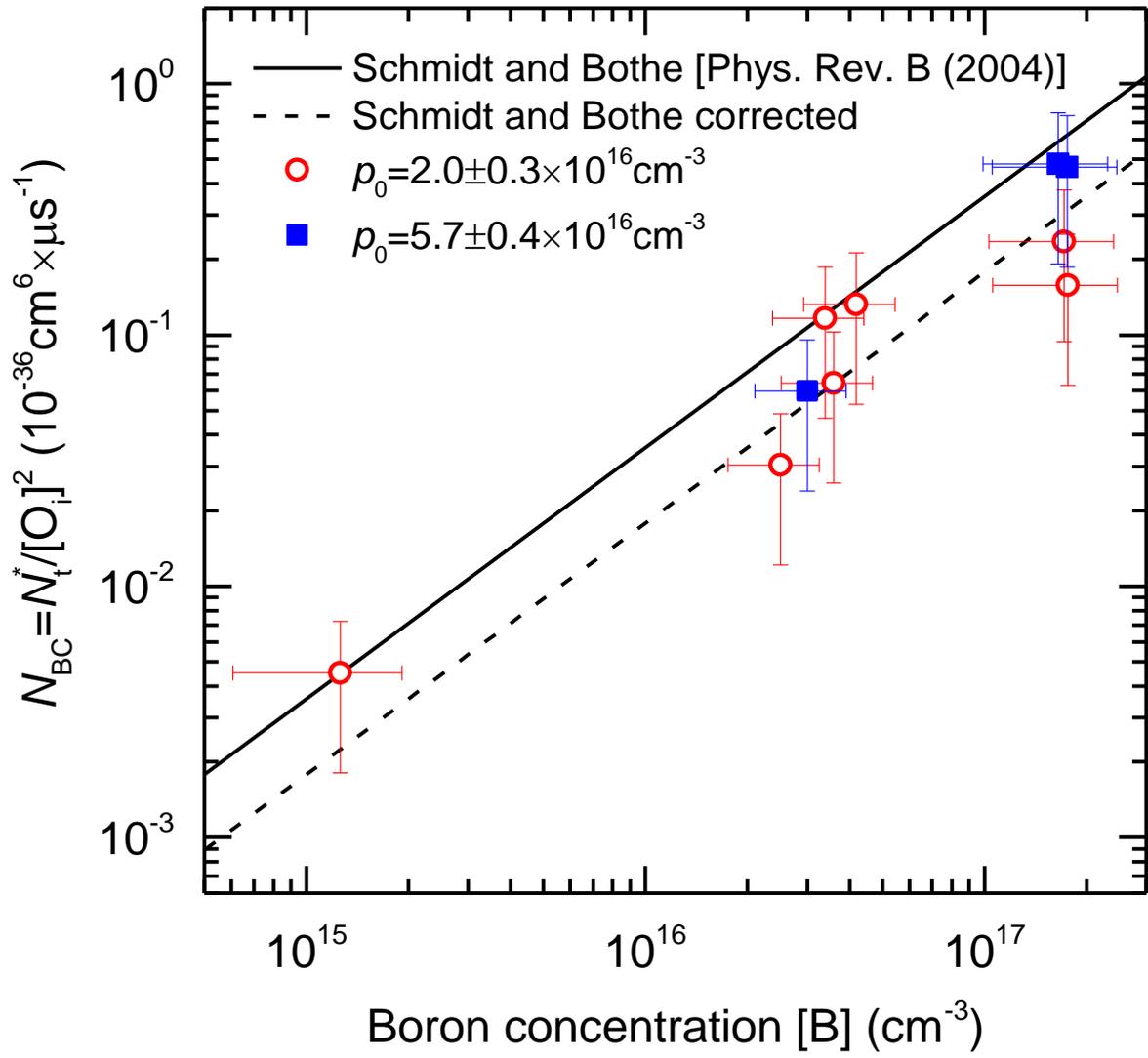


Figure 1 b

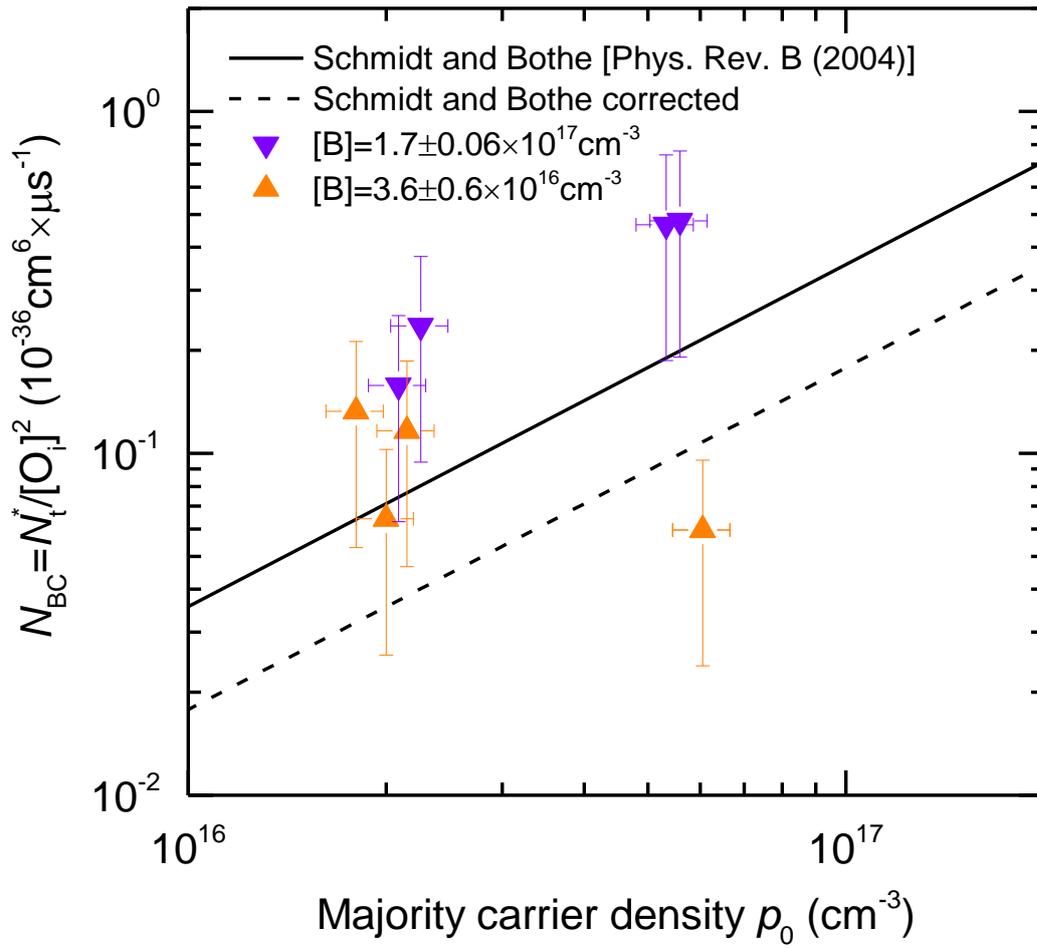


Figure 2 a

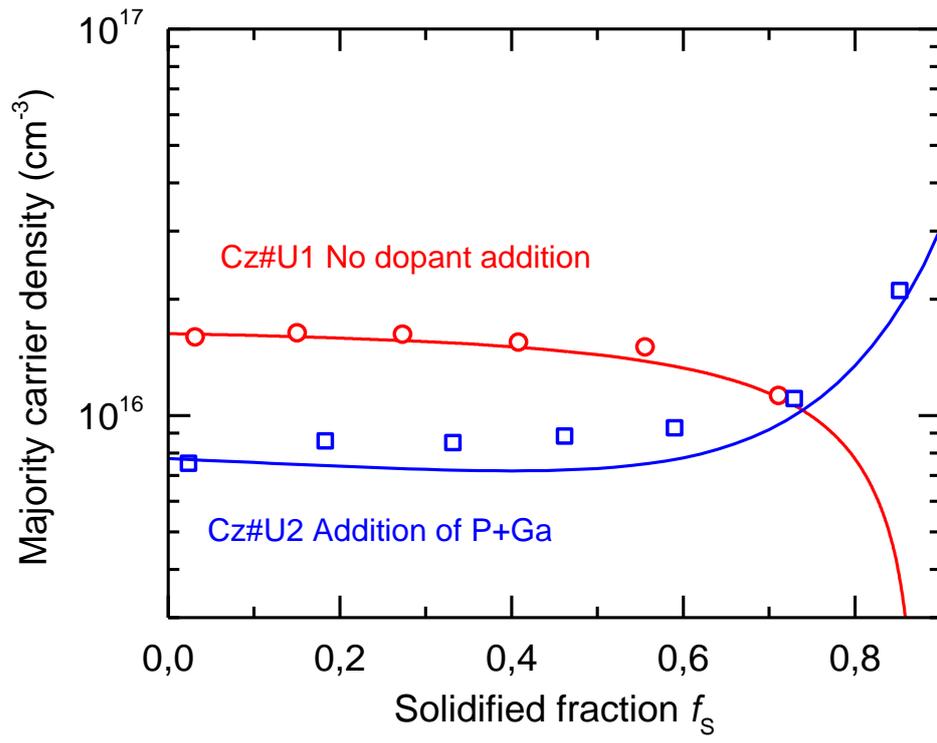


Figure 2 b

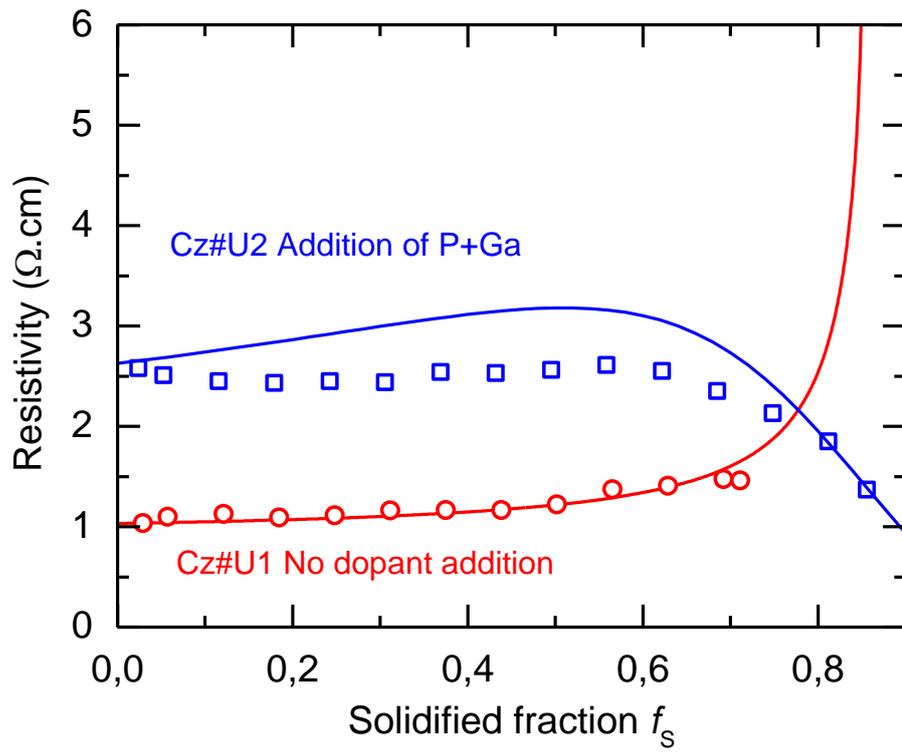


Figure 3

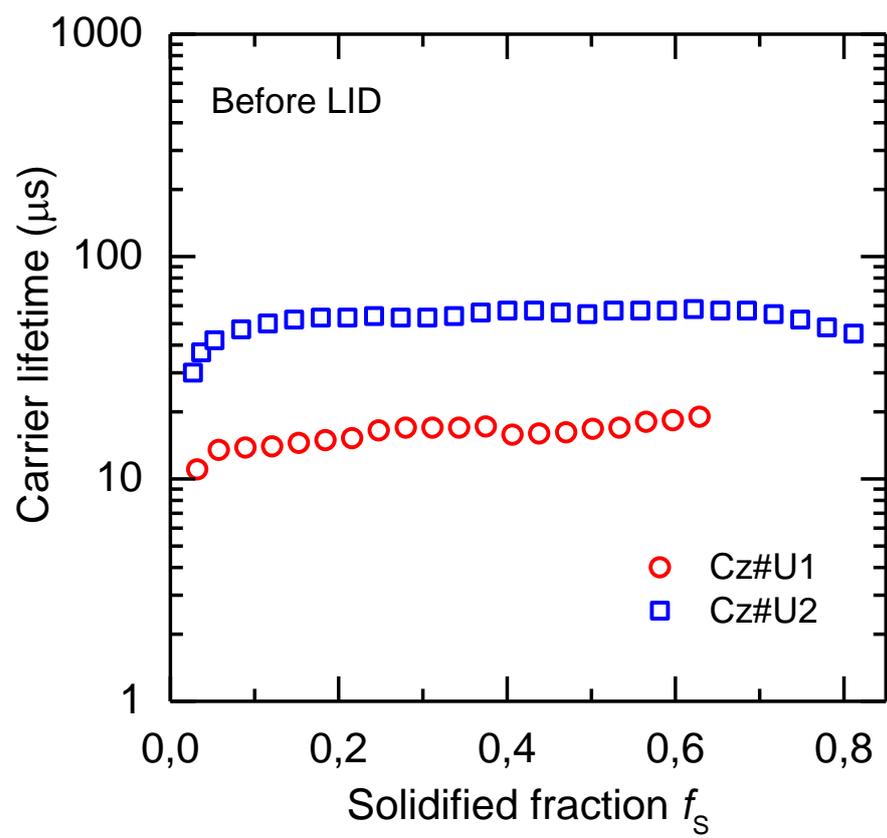


Figure 4 a

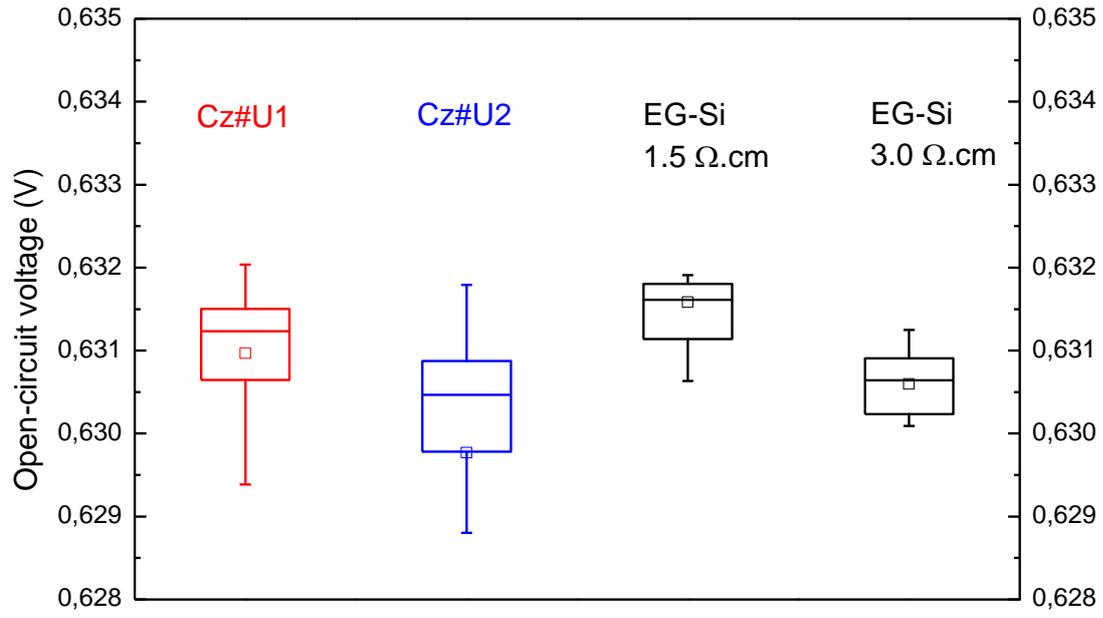


Figure 4 b

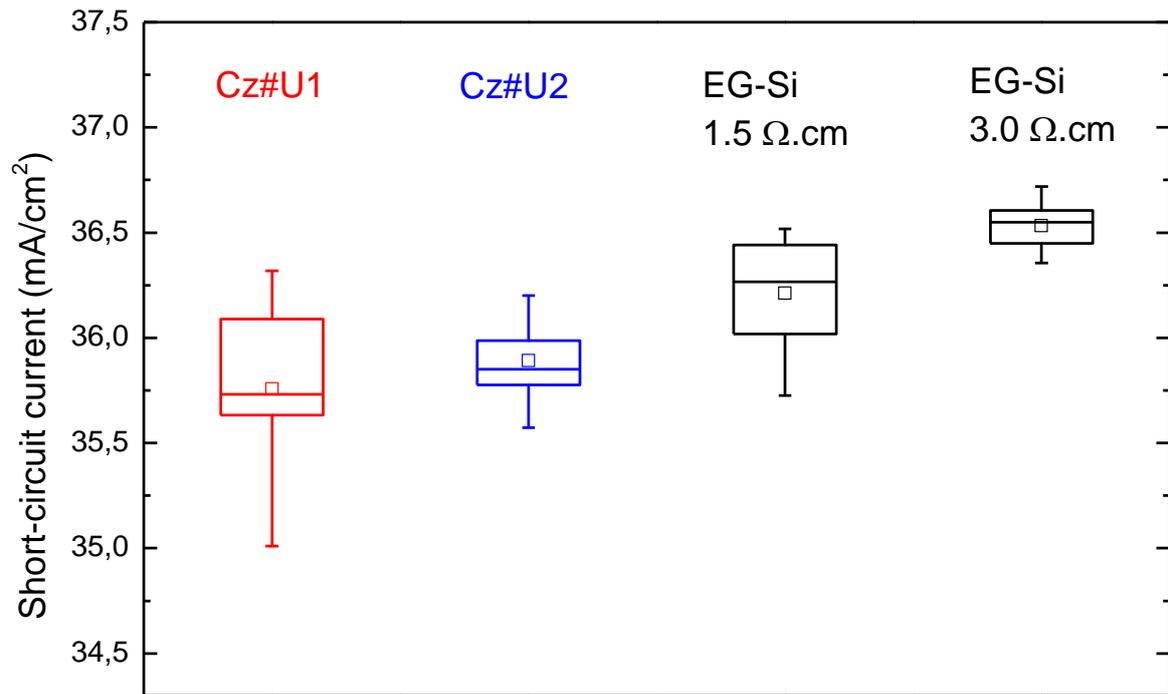


Figure 4 c

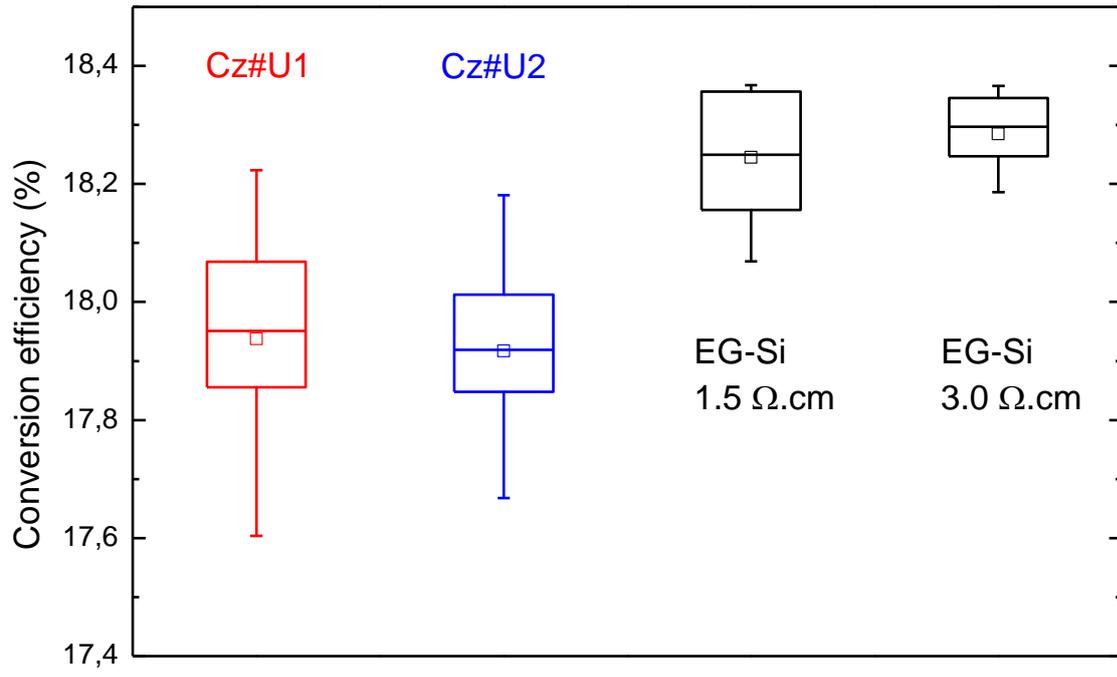


Figure 5

