Progress in Silicon Materials:

From Microelectronics to Photovoltaics and Optoelectronics

Edited by Deren Yang

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Preface

This proceedings volume contains the contributions of the Sino-German Symposium on "Progress in Silicon Materials: From Microelectronics to Photovoltaics and Optoelectronics", which was held in the Lily Hotel, Hangzhou, People's Republic of China from June 7th to 11th, 2002. 12 German and 13 Chinese scientists were invited to present their views about the different aspects of the silicon materials.

Microelectronics based on silicon was acting as a strong engine for the high-speed development of economy, science and technology in the world during last century. The steel age has given way to the silicon age and crystalline silicon was, is and will also be the main semiconductor material for the next decades. IC industry was continuously driving the progress for better and larger monocrystalline silicon wafers. Moreover, the world's production of solar cells in photovoltaics industry, which is dominated by low-cost crystalline silicon, has shown an annual growth rate of more than 20% during the past decades. Recently, the entire area of solar-grade silicon wafers has become comparable with the area of electronic-grade monocrystalline wafers for microelectronics. Furthermore, there are intensive research activities for developing optoelectronic devices based on silicon.

Silicon, one can reasonably argue, has already been the most intensively investigated material in the world. However, there is still a need for the continuation of research and development, which is the field of many laboratories all over the world.

In this symposium, the progress regarding crystalline silicon materials used for microelectronics, photovoltaics, and optoelectronics has been discussed with special attempt to reveal the synergy between these different areas. Both German and Chinese scientists from universities, research institutes and industries presented their views about the various facets of the silicon material from growth via processing to device related aspects. We hope that the symposium can provide a forum for a fruitful, intensive and open communication allowing the scientists to build up the relationship for a close cooperation in the future.

Finally, we would like to thank the Sino-German Science Center for its financial support and all the participants of the symposium for their contributions.

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Contents

Influence of Contamination on the Electrical Activity of Dislocations in Silicon

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ABSTRACT It is shown by EBIC that the dislocation activity is controlled by the degree of contamination with metals. The activity can be affected by external gettering and hydrogenation. DLTS investigations showed that occurrence and features of the dislocation-related C1 line depend on the amount of contamination, too. At a small amount, the impurities are believed to be tightly bonded to the dislocation core, resulting in a rather sharp energetic distribution of levels of the C1 line. At larger concentrations, the impurities are accommodated as a cloud in the dislocation strain field, giving rise to a broad energetic distribution of levels of C1. It is suggested that only impurity atoms in the cloud can be removed by external gettering and/or passivated by hydrogenation.

KEYWORDS dislocations, electrical activity, silicon, gettering, passivation

1. INTRODUCTION

Extended crystal defects in Si, such as dislocations or precipitates, may show a detrimental influence on the device performance due to their electrical activity. Thus generation of process-induced defects in the active device region has to be avoided. On the other hand, during microelectronics processing crystal defects are often formed intentionally as preferential gettering sites far away from the devices, with the purpose to attract metal impurities and keep them in the wafer interior (internal gettering). However, the electrically active defects in the wafer bulk cause a diffusion-current that may enhance the leakage of reverse-biased devices at elevated temperatures[1,2]. This disturbing electrical action of the defects/gettering sites in the wafer interior was found to increase with their

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gettering action[3]. Accordingly, understanding the influence of impurities on the electrical activity of crystal defects is still of interest for microelectronics. Recently, increased attention to the issue of defect activity was triggered by Si photovoltaics. The use of low-cost multicrystalline Si for solar cells demands alternative strategies for defect engineering since crystal defects/dislocations are inherent in this material. Improved knowledge about the interaction between impurities and extended defects is essential for further progress in this area.

In this paper we will discuss characterization techniques and findings related to the electrical activity of dislocations. After a brief review of our results as found by the technique of electron-beam-induced current (EBIC), we will give a detailed report on deep level transient spectroscopy (DLTS) investigations. It will be shown that the DLTS results allow a deeper insight into the accommodation of impurities at/near dislocations and a better understanding of the limits of gettering and hydrogenation treatments.

2. EBIC

The recombination properties of extended crystal defects in Si are mainly defined by recombination-active impurities decorating the particular defects, i. e. the recombination activity is not an intrinsic defect property, but of extrinsic origin. Even chemical-mechanical polishing[4] may cause defect contamination and increase the defect activity. Temperature dependent EBIC investigations of the recombination activity of dislocations clearly proved the role of defect contamination[5]. Both the magnitude of the contrast and its temperature dependence c (T) have been found to depend on the amount of contamination, see Figure 1. The c (T) behavior represents a fingerprint characterizing the degree of contamination of the crystal defects. Clean dislocations exhibit only very weak activity (type Ⅱ), with a maximum at about 50 K and untraceable activity at room temperature. Weak contamination leads to an increase of the low temperature activity, still leaving the room temperature activity below detection limit (type 2). Upon further increase of contamination the type of the temperature dependence changes and defect activity is detected at room temperature, too (type 1). A similar influence of contamination was observed for defects formed by oxygen precipitation in CZ-Si[6]. Finally, dislocations decorated with metal silicide precipitates have the highest activity throughout the whole temperature range (type Ⅰ). Internal Schottky junctions at the metal silicide particles that attract minority carriers and facilitate fast recombination are the cause of the strong activity[7].

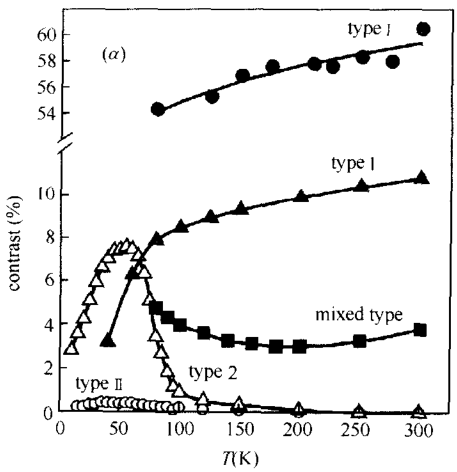

Figure 1 Temperature dependence of EBIC contrast of dislocations, c(T), for different concentration of contaminating impurities. Change of contrast type in the following sequence with increasing contamination: Ⅱ (clean) ->2->mixed->1->I

Experiments with hydrogenation and phosphorus diffusion gettering have shown that both passivation and gettering of impurities decorating dislocations change the c(T) behavior in the reverse sequence, from type 1 to type 2. However, the very small intrinsic activity (type Ⅱ behavior) can not be restored. The type 2 behavior has been found to represent the lower limit of activity that may be achieved by gettering and hydrogenation[8,9], corresponding to about $10^4 \sim 10^5$ deep levels/impurities per cm dislocation length (see below).

A model describing the dislocation c(T) behavior gives quantitative access to the concentration of the deep-level impurities at dislocations[10]. It delivers the following numbers of impurities per length unit for the types of behavior denoted in Figure 1: type Ⅱ-clean (no impurities), type 2-about $10^4 \sim 10^5$ impurities per cm dislocation length, type $1 \sim 10^6$ impurities per cm or more (for the mixed type behavior the number of impurities is between those for types 1 and 2). The model assumes that shallow 1-dimensional dislocation bands (about 80 meV from the band edges), induced by the dislocation strain field, and deep electronic levels, caused by impurity atoms segregated at the dislocations, can exchange electrons and holes. That is because the wave functions of impurity atoms-which are spatially located at or near the dislocation core in a distance not exceeding a few nm-and the 1-dimensional dislocations bands overlap. As a consequence, the recombination of carriers captured at the 1-dimensional dislocation bands can be drastically enhanced by the presence of even small concentrations of impurity atoms at the dislocation.

3. DLTS

By using DLTS we analyzed misfit dislocations in Si-SiGe (2%) stacks of n-type conductivity which were used as a kind of, model defects. The 60° misfit dislocations were located in a depth of about 2.3 μm below the surface. Samples with initially clean dislocations were contaminated with different amounts of gold, giving rise to either type 2 or mixed type/type 1 EBIC c(T) behavior. In samples with clean dislocations (type Ⅱ behavior) DLTS revealed one peak of an electron trap, MF.E1. In the intentionally contaminated samples with dislocations of type 2 or mixed type/type 1 behavior, respectively, we observed three electron traps, MF.E1, MF.E2 and MF.E3 (see Figure 2). We have clear evidence that only MF.E3 is due to the misfit dislocations, for details see Ref. 11. Arrhenius plots of the dislocation-related trap MF.E3-delivering an apparent enthalpy in the range between 0.39 and 0.52 eV-were found to be very similar to the plots of C1 line as observed by Omling et al.[12] and Cavalcoli et al.[13], for example. Accordingly, we believe that the electron trap MF.E3 is identical with the well-known C1 line.

Note that the C line is the most prominent dislocation-related line (in n-Si). It should be pointed out that this line could only be observed for contaminated dislocations. Clean dislocations did not show this line, see also Ref. 14.

Compared to the DLTS peak of point defects the C line is considerably broadened. On variation of the duration of the filling pulse, t_p, different behavior of the C line was reported in literature. The line shape was observed to be either symmetrically or asymmetrically broadened[13,15].

We found that these differences are related to the degree of contamination. We observed that the differing amount of contamination of the dislocations (with

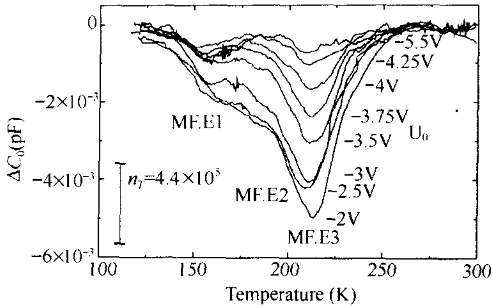

Figure 2 DLTS spectra taken with varying filling pulse height U_0 at pulse length $t_p=1$ ms, reverse bias $U_R=-8$ V, and emission rate is 100 s^{-1}. The axis ΔC_0 describes the absolute change of the capacitance which is directly proportional to the number of states. *Contaminated dislocations exhibiting type 2 behavior contain three electron traps* MF. E1, MF. E2 and MF. E3, respectively. Only MF. E3 is due to the dislocations

type 2 or mixed type/type 1 behavior, respectively) affects the t_p dependence of the C1 line and the energetic distribution of the dislocation-related states. To find out whether the distribution is sharp or broad we used a mode of measurement that benefits from the well-defined depth location of the misfit dislocations in our samples (at the interface between Si cap and SiGe layer): The height of the filling pulse, U_0, was kept constant while spectra were taken at different values of the reverse bias, U_R, applied during the emission cycle of electrons from the dislocation-related states. The larger the bias U_R the higher is the band bending, i. e. the deeper states can take part in the process of emission of electrons. Consequently, if there is a broad distribution of states we would expect that the maximum of the DLTS peak shifts to higher temperatures when U_R is increased (assuming that the capture cross-section of the states is nearly constant).

3.1 Dislocations with type 2 EBIC behavior

The C1 line was observed to remain symmetric during variation of the filling pulse t_p and the peak position was found not to shift on the temperature scale, see Figure 3a. Using the special mode of measurement described above, we found that the position of C1 line remains nearly constant when the reverse bias U_R is varied, see Figure 3b. There is a very small shift of the peak from 215 K at $U_R=-8$ V to

210 K at -4 V, only. Accordingly, we conclude a rather sharp energetic distribution of the dislocation-related states.

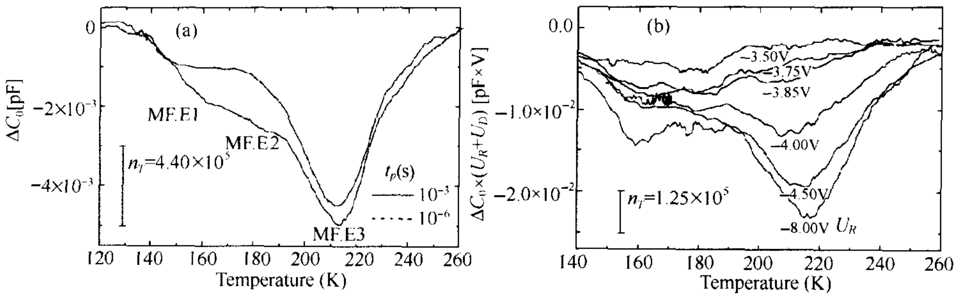

Figure 3 DLTS spectra of dislocations exhibiting type 2 behavior

(a) Dependence on the duration t_p of the filling pulse. Fixed parameters are height of the filling pulse $U_0=-2$ V, reverse bias $U_R=-8$ V, and emission rate is 100 s^{-1}

(b) Dependence on the reverse bias U_R. Fixed parameters are height of the filling pulse $U_0=-2$ V, duration of the filling pulse $t_p=1$ ms, and emission rate is 100 s^{-1}

This mode of measurement benefits from the well-defined depth location of the dislocations. The scaling of y axis, $\Delta C_0\times(U_R+U_D)$, takes into account the influence of the reverse bias and of the diffusion voltage. This allows to compare directly the absolute number of recharged levels. The measurements yield a rather sharp energetic distribution of dislocation-related levels for C1 line

3.2 Dislocations with mixed type/type 1 EBIC behavior

In this sample the contamination of dislocations is stronger. We observed that the C1 line becomes more and more asymmetric upon increase of t_p, while the position of the line maximum remains constant, see Figure 4a. Measurements using the special mode with variation of U_R, showed a significant shift of the maximum of the DLTS peak, see Figure 4b. The peak shifts from 185 K at $U_R=-3.5$ V to 212 K at $U_R=-8.0$ V. This means a broad energetic distribution of the dislocation-related states, with a width of about $\Delta E=50$ meV.

3.3 Number of states

An estimate of the absolute number of states related to the C1 line, n_T, as observed by DLTS under the Schottky contact of 1mm diameter gives for type 2 dislocations a few 10^5 states and for dislocations with mixed type/type 1 behavior a few 10^6 states, i.e. the difference between the samples amounts to about one

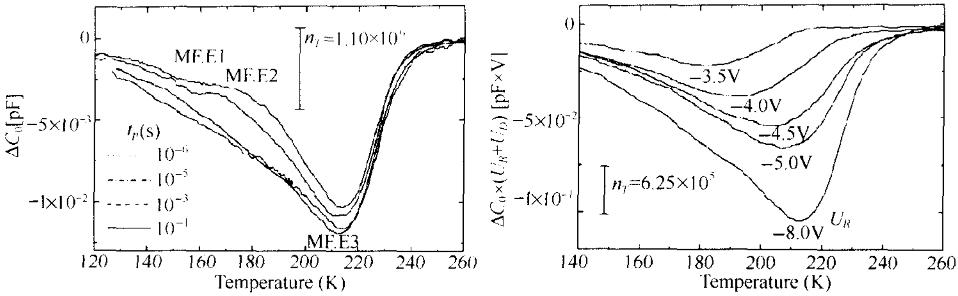

Figure 4 DLTS spectra of dislocations exhibiting mixed type/type 1 behavior

(a) Dependence on the duration t_p of the filling pulse. Fixed parameters are height of the filling pulse $U_0 = -2$ V, reverse bias $U_R = -8$ V, and emission rate is 100 s^{-1}

(b) Dependence on the reverse bias U_R. Fixed parameters are height of the filling pulse $U_0 = -2$ V, duration of the filling pulse $t_p = 1$ ms, and emission rate is 100 s^{-1}. The measurements yield a broad distribution of dislocation-related levels ($\Delta E = 50$ meV) for C1 line

order of magnitude. The total dislocation length under the contact was estimated to be in the range of 10 cm for both samples. Consequently, the state density per cm dislocation length is calculated to be a few 10^4 or a few 10^5 per cm, respectively. This is in a good agreement with estimates of the dislocation contamination as found by EBIC.

4. DISCUSSION

By using standard DLTS we observed that the position of the C1 maximum remained constant upon variation of t_p (see Figure 3a and 4a). The fact that the position of a DLTS peak stays constant upon variation of t_p is an indication that this peak is formed by a distribution of levels and not by band-like states. So we conclude that C1 is due to levels which are presumably caused by impurities. Indeed, the number of impurities as observed by EBIC agrees with the number of dislocation-related levels found by DLTS. According to the EBIC model[10] the impurities must be accommodated very close to the dislocations in a distance not exceeding a few nm.

The width of energetic distribution of the levels forming the C1 line increases with the degree of contamination or number of levels, respectively. It is found to be rather sharp for dislocations with type 2 behavior (Figure 3b) and to become broad/smeared out ($\Delta E = 50$ meV) for the dislocations with mixed type/type 1 behavior (Figure 4b), exhibiting a higher contamination density.

The reason for this dependence might be explained by a simple picture. We assume that the level and the related properties of a point defect/impurity atom are influenced by the configuration of the surrounding Si atoms it can see. This means the point defects/impurities at dislocations are exposed to local potentials, whereby their levels will be modified as compared to isolated point defects/ impurity atoms in the regular Si lattice. The surrounding of point defects/ impurities accommodated in the dislocation core differs strongly from the regular Si lattice. Point defects located in a cloud formed by the dislocation strain field see a surrounding which becomes more and more similar to the regular lattice with increasing distance from the core.

So, one may imagine a rather sharp distribution of levels of the impurities accommodated in the core region, differing strongly from that of the same species located in the regular lattice. Contrary, the point defects located in the cloud formed by the strain field may generate a broader distribution of levels. Inside the cloud point defects close to the core region are more strongly affected than the impurities located in the tail of the strain field. The latter are expected to form levels similar to isolated impurities in the regular lattice. Hence, the broad distribution of levels formed in the cloud may range between core-mediated impurity levels and levels being similar to isolated impurities.

If the number of point defects/impurity atoms is small enough there is a high probability that they are accommodated mainly in the dislocation core. In that case one expects a rather sharp distribution of levels as observed for dislocations with type 2 behavior (few 10^4 levels per cm). When the degree of contamination increases a large fraction of point defects may be located in the cloud formed by the strain field, extending a few nm from the core. Consequently, one may expect a broad distribution of levels, as in fact observed for dislocations of mixed type/ type 1 behavior (few 10^5 levels per cm).

Limit of action of passivation and external gettering

The outlined picture may be also helpful to explain the observed limits of gettering and passivation treatments. EBIC investigations have shown that a deep level density of $10^4 \sim 10^5$ per cm dislocation length remains active after passivation and gettering[8,9]. Using the above arguments we can understand this experimental observation as follows, see also Figure 5: (i) Impurities accommodated in the cloud surrounding the dislocation are weakly bound and can be passivated by hydrogen and/or can be removed by (phosphorus diffusion) gettering. (ii) Contrary, impurities accommodated in the dislocation core are strongly bound. We believe that this is the cause why the core-mediated impurity levels can neither be affected by hydrogenation nor removed by external gettering sinks.

Indeed, from our DLTS observations we concluded that only a few 10^4 impurities per cm dislocation length are accommodated in the dislocation core, and that the impurity atoms will be accommodated in the dislocation cloud for higher densities of contamination. It should be pointed out that simulation and further experimental work will be needed to prove the above suggestions.

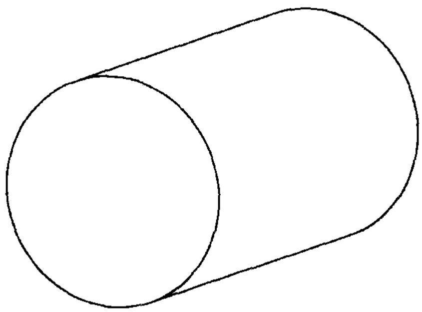

Figure 5 Suggested behavior of impurities accommodated at a dislocation (schematically). Note, the cloud of impurities is formed in the dilatation field of the dislocation. For simplicity we show here a circular distribution around the dislocation. Impurity atoms in the dislocation core are tightly bonded and cannot be affected by external gettering and hydrogenation while those atoms accommodated in a cloud surrounding the dislocation can be removed by external gettering and passivated by hydrogen

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