The Influence of High-Power Ion Beams and High-Intensity Short-Pulse Implantation of Ions on the Properties of Ceramic Silicon Carbide

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Abstract. The paper is focused on the study of the structural, electrical and optical characteristics of the ceramic silicon carbide before and after irradiation in the regimes of the high-power ion beams (HPIB) and high-intensity short-pulse implantation (HISPI) of carbon ions. The dominant mechanism of transport of charge carriers, their type and the energy spectrum of localized states (LS) of defects determining the properties of SiC were established. Electrical and optical characteristics of ceramic before and after irradiation are determined by the biographical and radiation defects whose band gap (BG) energy levels have a continuous energetic distribution. A dominant p-type activation component of conduction with participation of shallow acceptor levels 0.05–0.16 eV is complemented by hopping mechanism of conduction involving the defects LS with a density of $1.2 \times 10^{17} - 2.4 \times 10^{18}$ eV$^{-1}$·cm$^{-3}$ distributed near the Fermi level. The effect of radiation defects with deep levels in the BG on properties change dominates after HISPI. A new material with the changed electronic structure and properties is formed in the near surface layer of SiC after the impact of the HPIB.

1. Introduction
Silicon carbide SiC is a wide-band semiconductor material promising for high-power, high-temperature and radiation-resistant electronic devices. That stimulates study of the influence of the various kinds of radiation on the characteristics of SiC and of devices based on it [1–3]. SiC has a high chemical and mechanical resistance [3–5]. The band gap (BG) of SiC depending on the polytype varies in range 2.83–3.23 eV, and the threshold energy of defect formation it is 25–35 eV, that defines the high radiation resistance of the material. The characteristics of localized states in the BG of radiation defects (RD) induced by high-energy particles and their effect on properties of material depend on the type of particle and on the mode of irradiation [1–3]. The study of the effect on the properties of RD in the ceramic SiC is difficult and leads to ambiguous results owing to its complex structural hierarchy and high content of impurities and biographical defects (BD). Improvement of...
electrical characteristics of the epitaxial layers of SiC in devices is achieved by the ion implantation
and subsequent thermal annealing owing to an annihilation and redistribution of RD and modify of the
structure of the material surface [4–7]. High-intensity short-pulse implantation (HISPI) of ions
followed by heating of the surface layer provides simultaneous annealing of RD [8, 9].

The aim of the work is to study the structural, electrical and optical characteristics of ceramic SiC
before and after irradiation in the regimes of high-power ion beams (HPIB) and HISPI of carbon ions
C⁺ and to establish the reasons of the properties change.

2. Experimental

The investigations of morphology of the ceramic surface before and after irradiation by methods
described in [8] were performed by scanning electron microscopy. The elemental composition was
studied by energy dispersive microanalysis. Raman spectra were studied by using spectrometer
Nanofinder (λ=532 nm). The dark surface conduction σ and photocurrent dependence of the absorption coefficient
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and U<0 indicates a weak influence of the space charge on the dependence σ(U). Annealing until T=700–800 K increases the value of σ in 2–3 times due to the redistribution of charge carriers between shallow (ε<0.1 eV) or more deep traps (ε≥0.1 eV). Ceramics has a weak photosensitivity K=0.01 due to the influence of the BD with a high concentration of N>10^{16} cm^{-3} distributed on the grain boundaries. Then–type of σB and odominates, as photo and thermostimulated currents I_{PHS}(T_{SCE}(T)) show. Influence of the donor BD having the impurity or vacancy nature on the transport of charge carriers prevails. Temperature dependences of σ, σB(T) within the interval T=300–700 K are determined by the thermally stimulated electron exchange between shallow donor levels with the activation energies ε_{α1}=0.06–0.061 eV, ε_{α2}=0.2–0.21 eV, ε_{α3}=0.4–0.43 eV and conduction band (CB) (figure 2). Population of traps by charge carriers is redistributed between the single levels of energy ε_{α1,3} after sequential heating owing to low values of their degree of population n/N=10^{-10}–10^{-6}. The factor σB indicates that (n/N)_{CB}>(n/N)_{α1}>(n/N)_{α3}. The values ε<sub>α</sub> below their values for intrinsic defects of SiC [1–3] due to the interaction between BD and a continuous distribution of LS in the BG, by analogy with [10, 11]. The shallow donor levels with ε<sub>α</sub>=(0.1–0.3) eV may be due to impurity atoms N and Ti or vacancy V_{Si} in [1, 2]. Oxygen atoms (content<5.5 at. %) in complexes with BD as well as Vatoms have a deep levels (ε=0.65–1.59 eV [2]), which impact on reduce of values of σ.

The electrical parameters ε<sub>α</sub>, σB, K, n/N indicate the effect on transport of the hopping conduction that is verified by approximation σ(T) by the Eq. (2) within the interval T=300–700K. LS density with the participation of which carried out σB(T) is N(E_F)=(2–9)×10^{17} eV^{-1}·cm^{-3}. The value of the most probable jump distance calculated according to [10] with the values N(E_F) is R=7–11 nm. The electronic structure of materials such as ceramics can be described in frame of the model with the function N(ε), which varies little with depth of LS in BG [10, 11]. The sharp edges of the valence band (VB) and CB as in crystals do not exist. Fermi level in SiC is pinned near the middle of the BG at ε=1.5 eV due to the high concentration of defects and impurity atoms O, V, Ti. This is confirmed by low values of σ≤2×10^{-4}S (figures 1 and 2) and K≤0.01 and by localization of a strong bands at energies ε=1.4 and 1.7 eV in the absorption spectra α(hv) (figures 1 and 2). Despite the dominance of n–type of σ and σB we cannot exclude the influence of acceptor defects on the properties.

Irradiation of SiC significantly alters the morphology and structure of the surface and thin near surface layer (until 200 nm), which is reflected in CVC, σ(T), in change of the type of σ (n→p) and in parameters ε<sub>α</sub>, σB, T, σB, N(E_F), R and in spectra α(hv) (figures 1–3). Influence of HPB and HISPI on properties change is different (figures 1–4). Changes of the elemental composition of SiC after HPB and HISPI showed that increase of the content of silicon from 20 to 40 at. % has the greatest effect on the growth of conduction (figure 3). At the same time the content of carbon in layers of SiC decreases from 80 to 40 at. % and oxygen decreases from 6 to 1 at. % (figure 4). A similar correlation between the content of C, Si and concentration of the donors was fixed in 6H–SiC (n–type ofσ) [2]. HPB significantly increases σ and changes its characteristics (figures 1 and 2). CVC becomes almost linear (s=0.99–1.05, a/b=0.01–0.02) (figure 1). The depth of the shallow centers decreases from ε_{α1}=0.06 to 0.034 eV and from ε_{α2}=0.21 to 0.123 and 0.155 eV and values of (n/N)_{α1,2} increases in 10^{2–3} times as shown a factor σB. The deep centers with ε_{α3} do not appear. Currents I_{PHS}(T) and their ε_{α} indicate the predominance of exchange between holes and VB and LS with ε=ε_{α1}+ε_{α1,2} (ε_{α} is the top of VB). The manifestation of stable defects V<sub>C</sub>–C<sub>Si</sub> is more probable in SiC with p–type of conduction while the defects V<sub>Si</sub>^{(2−)} dominate in case of n–type of σ[1, 2]. Shallow acceptor levels are associated with impurity of Al (≤0.09 at.% whose energetic levels at ε=ε_{α}+(0.1–0.3) eV activated by interaction with RD[2]. I_{PHS}(T)<0 show that deep donor levels with ε<1.5 eV have the effect on parameters also.LS density for hopping conduction increases from N(E_F)≈(2–9)×10^{17} to (1.2–2.4)×10^{17} eV^{-1}·cm^{-3} and jump distance decreases from R=7–11 to 5–7 nm, which means increasing the role of σB. The surface morphology change occurs due to its melting and the impact of nanoparticles of Si, C and metals. Partial annealing of RD and their association in a complex with BD occur owing to heating of surface.
After HISPI σ decreases or not changes and nonlinearity of CVC increases (s=1.2–2.3, a/b=2–5) (figure 1). This is caused by the accumulation of RD with deep LS that are centers of trapping of charge carriers. Parameters of RD with εσ1 are close to their values prior to irradiation and of centers with εσ2 to parameters levels induced by HPIB. Deep centers with εσ3 appear also. Effect of donors increases, as shown I_{TSC(T)}. LS density for realization of σp is N(E_F)=(1.2–11)∙10^{17} eV^{-1} cm^{-3} at R=6–12 nm. LS density decreases with σ increasing. Comparison of parameters of σp and σa shows that the influence of BD after HISPI is more greatly than after HPIB. Interrelation between electrical parameters allows to clarify mechanism of transport charge (figure 4).

Value of α decreases in 2–5 times after HPIB while α not changes or increases in 1.2–1.5 times after HISPI. Shape of α(hv) after HPIB changes similarly as in SiC after a powerful laser irradiation[6]. Absorption band at 1.4–2.8 eV induced by BD (concentration N=(2–20)∙10^{18} cm^{-3}) separates from the band of interband absorption at 2.8–3.3 eV characteristic for amorphous and highly

**Figure 1.** CVC of SiC before (curve 1) and after HPIB (curves 2 and 3) and HISPI (curves 4–6): J=2.4 J/cm², n=3 (curve 2), 50 (curve 3); J=0.2–0.3 J/cm², n=300 (curves 4–6).

**Figure 2.** Temperature dependencies σ(T) of SiC before (curve 1) and after HPIB (curves 2 and 3) and HISPI (curves 4–6): J=2.4 J/cm², n=3 (curve 2), 50 (curve 3); J=0.2–0.3 J/cm², n=300 (curves 4–6).

**Figure 3.** Effect of the elemental composition of SiC on its conduction σ before and after irradiation in the regimes of HPIB and HISPI.

**Figure 4.** N(E_F) vs. εσ in SiC before (x) and after irradiation (∗, ◆).
defective materials. Average value of \( BGE_{\alpha} = 3 \) eV is in limits of the values \( E_g = 2.83 \sim 3.23 \) eV peculiar to different polytypes of SiC [6]. Conservation of this band after HISPI indicates a negligible change of the absorption edge. Accumulation of the RD levels after HISPI strengthens the pinning of \( E_F \) near the middle of BG. There are three groups of bands with centers at \( \varepsilon_1 = 1.4 \) eV; \( \varepsilon_2 = 1.72 \) eV, \( \varepsilon_3 = 2.4 \) eV. The concentration of these centers before irradiation is: \( N_1 = (2 \sim 2.3) \times 10^{18} \) cm\(^{-3}\), \( N_2 = (5 \sim 7) \times 10^{19} \) cm\(^{-3}\); \( N_3 = (6 \sim 7.3) \times 10^{18} \) cm\(^{-3}\). After HPIB \( N_1 = (1 \sim 1.2) \times 10^{18} \) cm\(^{-3}\), \( N_2 = (2 \sim 2.2) \times 10^{19} \) cm\(^{-3}\), \( N_3 = (2.5 \sim 2.7) \times 10^{18} \) cm\(^{-3}\). Taking into account [1, 2] bands with \( \varepsilon_{1,3} \) were identified with vacancy RD \( V_{C_{Si}} \). Single RD clearly manifest after HISPI. Transitions between the LS with a continuous distribution dominate after HPIB. Formation after HPIB of a new material with changing electronic structure and properties has an effect on dependencies \( \alpha(\sigma, \varepsilon_\alpha) \) also. Dependency \( N(E_F)(\varepsilon_\alpha) \) in SiC before and after irradiation is typical for materials containing a high concentration of defects (figure 4) [10]. Parameters \( N(E_F) \) and \( \varepsilon_\alpha \) are distributed on two arrays for shallow and more deep centers (figure 4). Decrease of LS density from \( N(E_F) = 2.4 \times 10^{19} \) to \( 1.2 \times 10^{17} \) eV\(^{-1}\)·cm\(^{-3}\) correlates with the increasing of absorption coefficient from \( \alpha = 3 \times 10^{6} \) to \( 10^{4} \) cm\(^{-1}\) and decreasing of \( \varepsilon_\alpha \) by analogy with [10]. The impact of \( \sigma_p \) on transport charge for values \( N(E_F) \) and \( \varepsilon_\alpha \) in band 1 is higher than in band 2 (figure 4).

4. Conclusion

Electrical and optical properties of SiC ceramic before and after irradiation are conditioned by BD and RD whose levels are continuously distributed on energy in BG. Defects and impurities with high concentration \((2 \sim 20) \times 10^{18} \) cm\(^{-3}\) are heterogeneously distributed along the boundaries between the structural fragments of ceramic. The activation component of the p-type conduction after irradiation is realized in the exchange of holes between the VB and shallow acceptor levels with the activation energy of \( \varepsilon_\alpha < 0.3 \) eV and is complemented by the hopping mechanism of transport on LS near the Fermi level which localized in the middle of BG. Density of LS for realization of hopping conduction is changed after irradiation from \((2 \sim 9) \times 10^{17} \) to \((1.2 \sim 24) \times 10^{17} \) eV\(^{-1}\)·cm\(^{-3}\) and impact of hopping component on electrical parameters is amplified. Influence of RD with deep levels on the optical and electrical properties of SiC dominates after HISPI. A new material with the changed electronic structure and properties is formed in the near surface layer of SiC after the impact of the HPIB.

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