Damage range in swift Xe^{26+} ion-irradiated polycrystalline iron and silver studied by positron annihilation technique, long-range effect

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Abstract

Investigations of defects and their spatial distribution in iron and silver irradiated with swift Xe^{26+} ions of different energies: 18.5, 45.5, 122.5 and 167 MeV have been performed using conventional positron lifetime spectroscopy. In the implanted layer, in which the ions are stopped, dislocations and clusters of vacancies were found in iron and only dislocations in silver. In both metals, an extended layer of damage with defects was detected beyond the implanted layer. The origin of this extended layer can be explained by the depth distribution of the stress caused by swift ions in the implanted layer. The calculations using a finite element method indicates that the stress value in the implanted layer, i.e., about 1 GPa for iron and 0.5 GPa for silver exceeds several times the yield strength.

1 Introduction

The interaction of swift ions with matter is widely studied because of the importance of both practical and cognitive perspectives [1]. The entry of fast electrons, protons or other ions causes their kinetic energy to be transferred to the target material. The energy is accumulated in a very short time, i.e., an order of picoseconds and this results in distribution of displacement of lattice atoms and/or generation of other defects. This is at the atomic level, but the physical properties also change in the macroscale.

The range of implantation of these particles depends mainly on their charge, velocity and target properties. The implanted layer that covers the layer from the entrance surface to the ion-projected range (R_p) is damaged. Its lattice structure contains displacement spikes, damage cascades, many type of point defects, dislocations and their loops. Due to the great impact of ion implantation this is the subject of many studies using various experimental techniques. These

facts have been described in numerous papers and books, e.g., Ref. [2].

However, in the literature, it was reported in addition that a layer containing a dense dislocation network extends beyond the implanted layer [3–5]. We will call it the extended damage layer. It may result from static and/or dynamic stresses caused by implantation, presence of the implanted ions and/or swelling. Surprisingly, there are reports that the thickness of this layer is two to three orders of magnitude larger than the thickness of the implanted layer. Another surprise is that the dislocation density instead of monotonically decays with the depth increase exhibits a maximum at a certain depth and then decays with the depth increase [2]. The occurrence of the damaged layer beyond the projected range is called a long-range effect (LRE) [5]. However, its nature is not sufficiently well explained and it seems that there is no consensus about its existence. Many experimental techniques have been used to study this effect, but rarely the spectroscopy of positron annihilation [3].

Positron annihilation techniques are extremely sensitive to the open volume defects, such as vacancies, their clusters or dislocations. That is due to the fact that the measured positron annihilation characteristics, i.e., the annihilation line shape or positron lifetime are sensitive to the local electron density in the annihilation site, see for instance Ref. [6]. The density disorder is reflected in these values. Moreover, positively charged positron is also sensitive to changes in the distribution of host ions as well. This can happen

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when open volume defects such as vacancies, their clusters or jogs at the dislocation line are present in the host. Such defects are traps for thermalized positrons and they are well recognized in the measured positron annihilation characteristic [7]. The above statements are valid because positrons before and during annihilation are in a thermal state, thus only electron momentum contributes to momentum of the annihilating pair.

The use of the positron annihilation technique for the implanted and extended damage layer studies is limited. Conventional positron techniques apply the β^+ isotopes as the positron source, e.g., ^{22}Na . The implantation range of such positrons is large, e.g., in iron it is about 26 μm , see “Appendix 1”, and this is more than two or three orders of magnitude greater than the typical thickness of the implanted layer [8]. Therefore, the variable energy positron beam (VEP) measurement seems to be a right tool for studies of damage in the implanted layer. In this measurement, the energy of implanted positrons is reduced to dozens of keV, and the implantation range is limited to a few micrometers or less [9]. Many authors applied this technique to the study of different materials exposed to the various ion implantations [6]; however, observation of LRE is not unambiguous.

Liszkay et al. performed VEP and XTEM measurements of the Si(100) irradiated with 18 keV He ions [10]. They observed voids at the depth of 180–240 nm and this corresponds with the projectile range equal to 190 nm. Similar good agreement between the two ranges was observed in VEP measurements of stainless steel exposed to proton radiation with different energies from 95 to 220 keV and dose from 10^{14} to 10^{16} ions/ cm^2 [11]. Nevertheless, Kinomura et al. observed the difference between the measured and simulated by the Stopping and Range of Ions in Matter (SRIM) code [12], defect ranges in the iron sample irradiated with 150 keV Ar^+ ions at low temperature of about 100 K [13]. They concluded that the origin of such differences cannot be explained simply by vacancy diffusion, poor accuracy of positron implantation depths and recoils from the surface impurities. Similar observation was done by Anwand et al. in the case of 6H-SiC, implanted with 65 keV Al^+ and 120 keV N^+ [14]. More sophisticated technique, i.e., variable energy pulsed positron beam was applied by Mazzoldi et al. to studies of silica glass irradiated with Ar^+ ions at 30 and 50 keV with different doses [15]. They also observed the distribution of defects at a depth twice as large as the projected range for such an ion.

There are two objections to the results obtained using the VEP technique with regard to LRE. First, the depth scaling must be done with the impact energy of the implanted positrons, and this relation is obtained from computer simulations. Authors use different codes, and this can lead to some discrepancies. Second, in description, the diffusion of thermal positrons of the VEP results must be taken also into

account. This can lead to some inaccuracies in scaling the depth from the entrance surface.

In our recent publications, we suggested another experiment for verification of LRE, which is using the conventional positron experimental technique. For scanning the defect depth profile from the entrance surface, the sequenced etching of implanted sample and measurements of positron lifetime were proposed, for details see “Appendix 2”. This allowed us to avoid the objections mentioned above. However, to do this, in our experiment swift Xe^{26+} ions of high energy equal to 167 MeV were implanted into well-annealed metals: iron [16], copper [17] and silver [18] and recently titanium [19]. Indeed, we observed some defects beyond the projected range obtained from SRIM code; however, their range was considerably shorter than those reported in the literature.

In this paper, we present systematic positron annihilation studies to find out more solid statement about LRE. For this purpose, the iron and silver samples were irradiated with Xe^{26+} ions of various energies. This should confirm the validity of the experimental techniques, which we used. The experimental results were compared with the theoretical calculations of stress distribution for the modeled system using a finite element method.

2 Experimental details

In our research we used samples of pure polycrystalline iron (99.99% of purity) and silver (99.8% of purity) purchased from Goodfellow with the dimension of $15 \times 15 \times 0.3$ mm^3 . All iron and silver samples were annealed at 1000 and 600 °C, respectively, for 2 h in the vacuum conditions of 10^{-5} Torr, for, before treatment. To remove possible oxides on the surface they were etched in the 25% solution of nitric acid. The measured positron lifetime in such samples showed only single lifetime equal to 109 ± 1 and 136 ± 1 ps for iron and silver, respectively. These values correspond to the bulk value in pure iron reported by many authors [7]. This indicates that in such virgin samples, only residual defects remain which cannot interfere with defects caused during ion implantation.

The samples were irradiated with Xe^{26+} ions at IC-100 cyclotron working at Flerov Laboratory of Nuclear Reactions at Joint Institute for Nuclear Research (JINR) in Dubna, Russia. The primary kinetic energy of ions is about 167 MeV; however, for its reduction thin aluminum foil as a beam degrader was placed on the sample surface. The average flux was 5×10^9 $\text{cm}^{-2} \text{ s}^{-1}$ and only one dose, i.e., 10^{14} ions/ cm^2 was applied for all samples. Temperature during irradiation was not higher than 80 °C.

For detecting damage caused by radiation, the conventional positron lifetime spectroscopy was applied.

Radioactive isotope ²²Na of about 1 MBq activity in the form of NaCl deposited on to a thin 7 µm Kapton foil was used. The positrons from the radioactive source are implanted into two identical samples sandwiched the source. The gamma photons emerging at the time of their emission (1.276 MeV) and annihilation (0.511 MeV) are detected by two radiation detectors, which were BaF₂ scintillators coupled with Philips XP2020Q photomultiplier tubes. The fast–fast gamma–gamma coincidence setup constituted by them together with the associated electronics had a time resolution of about 280 ps.

Computer simulations of ion trajectories were performed using SRIM code. In addition, the calculations of stress distribution were performed using finite element method with ABAQUS 2017 software [20].

3 Results and discussion

Samples of pure iron were irradiated with Xe²⁶⁺ energy of 167 MeV; however, the ions first passed through the aluminum foil, which acts as degrader of the primary ion beam to reduce ion kinetic energy. The thickness of the foil was equal to 3, 9.5 and 13 µm. Then the irradiated samples, without foil, were subjected to measurement of the positron lifetime spectrum and sequenced etching. Only about 2-µm-thick layer was removed from each sample in each step. At the initial steps, two lifetime components in the measured spectra were resolved; however, only a single component was observed deeper. In Fig. 1, all the results for three iron samples are depicted, i.e., the value of the first lifetime component τ_1 , on the top, below the value of the second lifetime component τ_2 and then the value of its intensity I_2 . The value of the first lifetime component can be calculated as follows: $I_1 = 100 - I_2$. On the bottom, the value of the mean positron lifetime is presented and calculated as follows:

$$\bar{\tau} = (I_1 \tau_1 + I_2 \tau_2) / 100\%. \quad (1)$$

This parameter is used to infer information on the cumulative defect evolution processes taking place within the samples while the experimental parameter is varied through systematical steps. This is a solid parameter that is not sensitive to the possible small fit ambiguity. In Fig. 1, the gray rectangle represents the region of the implanted layer. The projected range R_d for each sample was calculated using the SRIM code, and it is equal to 7.1, 4.2 and 2.3 µm, for the thickness of the aluminum foil equals to 3, 9.5 and 13 µm, respectively. In addition, in Fig. 1, the hatched region represents the value of the bulk positron lifetime equal to 109 ± 1 ps, which was obtained for the virgin samples before implantation. In our studies, we always assume that after etching the layer damage by implantation, the values of the

positron lifetime must return to this value. The depth, where it occurs, is the total depth of the damaged region.

In all cases in Fig. 1, the values of τ_1 reach the hatched region; however, it exceeds the depth of the R_d value (the right edge of the gray rectangle). The value of τ_1 in the implantation layer in all cases is higher than the bulk value; however, it is lower than the value of 176 ps which corresponds to the positron annihilation at a monovacancy in the perfect host. This indicates the presence of dislocations which contain jogs, being the positron traps. In the literature, according to theoretical calculations the value of about 117 ps corresponds to jogs on the edge dislocation, see Ref. [21, 22]. They are distributed also slightly beyond the implanted layer, i.e., the gray rectangular region, Fig. 1. The value of the second lifetime τ_2 ranges from 260 to 380 ps. This reveals the presence of large vacancy clusters which consists of more than nine vacancies [23]. The intensity of this component is the highest in the implanted layer; however, its value gradually decreases with the depth. This indicates that number of these clusters decreases with the depth and, what is more interesting, exceeds the implanted layer region. In the implanted layer, the intensity of second lifetime component is the highest, i.e., 5– to 27%, and beyond this layer the intensity is lower, less than 2%. The origin of the vacancy clusters in the implanted layer can be explained as a residue from depleted zones or spikes caused by the swift ions, but beyond where there is no ion pass, a different process should be considered. For instance, the interaction of dislocations can generate the vacancies which merge into clusters. Another feature can be indicated, the intensity I_2 at the entrance surface also depends on the thickness of the degrader foil, it is lower for the 13-µm-thick aluminum foil and then gradually increases with the reduction of the foil thickness. One can remind that similar values of the positron lifetimes and their intensities were obtained in iron samples exposed to the dry sliding, where only mechanical damage of the worn surface takes place [24]. However, the intensity of the long lifetime component at the surface was less than 10%. Also, in this case, the value of the second lifetime intensity decreases with the depth increase. The total damage depth is much deeper, i.e., about 300 µm.

The gradual decrease of the mean positron lifetime is clearly visible in all cases in Fig. 1, bottom figure. In Fig. 2, these values are re-drawn adding the values for which the degrader foil was not used (the values are taken from Ref. [16]). It can be seen that the thickness of the aluminum foil significantly influences the total depth of the damaged region. This is well understood because the increase of the degrader foil thickness causes the reduction of ion kinetic energy and velocity when they are directly implanted into the iron samples, and this causes the decrease of the projectile range. The energy of the ions entering the iron after passing the degrader foil is about 122.5, 45.5 and 18.5 MeV for

Fig. 1 The values of the components, τ_1 , τ_2 and intensity of the second component I_2 resolved from the measured positron lifetime spectra as the function of the depth from the entrance surface of iron samples irradiated with Xe^{26+} ions. The bottom figure shows the values of the mean positron lifetime calculated from Eq. (1). The results were

obtained for three values of the aluminum foil thickness, d_{Al} . The foil was used as a degrader of the primary ion beam of 167 MeV. The gray rectangle represents the implanted layer, which thickness was calculated from SRIM code. The hatched region represents the positron bulk value, obtained for the virgin sample

the thickness of aluminum foil equals to 3, 9.5 and 13 μm , respectively. Although this is an obvious conclusion, it supports the usefulness of the applied etching method to this type of studies. It is also markedly visible that the increase of the thickness reduces the value of the mean positron lifetime in the implanted layer. It means that reduction of the ion energy causes also less damage in the implanted layer. This is in contradiction to the theoretical calculations obtained using SRIM code.

On the top of Fig. 2, the depth profile of vacancies predicted using this code with the Kinchin–Pease option,

assuming a displacement threshold energy of 25 eV, is depicted for all cases. Only trajectory of 1000 ions was taken in the calculations. It is apparent that the thickness of the degrader foil does not affect the amount of vacancies in the implanted layer which is not reflected in the bottom figure with the experimental values. There is also no increase in the mean positron lifetime values at the end of the ion trajectory, i.e., near the end of the implanted layer, as it is visible in the top figure. It should be emphasized that in these results there is no Bragg peak, which is present in the stopping power depth dependency. These contradictions with experimental

Fig. 2 The depth distribution of the mean positron lifetime for iron samples irradiated with Xe^{26+} energy of 167 MeV after passing the aluminum foil of d_{AL} thickness tagged in the figure (bottom). The results were taken from Fig. 1; however, for the case of implantation without an aluminum foil, the results were copied from Ref. [16]. The top graph shows the distribution of vacancies in iron plate calculated using SRIM code for different d_{AL} and the corresponding projected range R_d in iron, see text

and theoretical calculations are well understood because the theoretical calculations do not take into account the evolution of defects caused by, for example, the temperature and stress. The theoretical predictions do not show also any defects beyond the projectile range R_d ; however, beyond the implanted layer the mean positron lifetime value is slightly above the bulk value, i.e., 115 ps at certain depths for all cases. This clearly indicates a certain amount of dislocations and/or open volume defects in this extended region. In this case, the thickness of the degrader foil does not affect the value of the mean positron lifetime. It indicates that the damage at the depth beyond R_d has a different origin than direct ion implantation.

The damage layer beyond the implanted layer is extended up the depth larger than R_d value, Fig. 1. For instance, in the case of the 9.5-μm-thick foil, the projectile range is 4.1 μm

but the total depth of the damage region is about 20 μm. This is an interesting feature that may be the aforementioned LRE, but at much smaller distance from the entrance surface. Our observations correspond rather to the results obtained using slow positron beam, where it was revealed that this second defect distribution extends more than two times deeper than R_d . Recently, Lu et al. reported similar effect, detected in TEM observation in irradiated by 3 MeV Au ions single crystalline Ni, NiCo and NiFe alloys [25]. These authors also observed the damage layer and defect cluster stretched to a depth of about 1 μm in panoramic cross-sectional TEM images whereas the R_d obtained from SRIM code was about 0.28 μm. This effect cannot be explained by the statistical nature of the ion trajectory, the straggling, which in the case of Xe^{26+} ions is about 0.4 μm. Additionally, Lu et al. have shown that the range of damage increases

with the increase of the dose of implanted Au ions. In our former studies such an effect was not observed clearly. In copper [17] and titanium [19] irradiated by 167 MeV Xe^{26+} ions applied dose was in the range from 10^{12} to 10^{14} ion/ cm^2 . It seems that this range could be too narrow to observe the effect, because the dose reported in Ref. [25] was in the range from 2×10^{13} to 5×10^{15} ions/ cm^2 .

In the case of pure silver irradiated with Xe^{26+} ion beam of 167 MeV such effect seems to be observed as well. In this case, only two values of dose were applied: 5×10^{12} and 10^{14} ions/ cm^2 . Single lifetime component was resolved in the measured spectra, which is depicted in Fig. 3. The measured values are below the value for monovacancy, i.e., 208 [7]. Thus, only dislocations with jogs are present and there is no vacancy cluster, as it is in iron. A striking feature is the fact that the bulk value of about 136 ± 1 ps (tagged by dashed region) was not achieved even at the depth of 35 μm . Thus, in silver the extended damage layer must be much thicker than in iron. This measurement was performed for the second set of silver samples; however, the first set exhibited identical feature, as it was reported in Ref. [18]. The value of the positron lifetime in the region directly beyond the implanted layer depends on the implanted dose, for the lower dose its value is lower than for the higher dose, Fig. 3. The explanation why the bulk value was not obtained after deep etching is given below.

Lu et al. suggest two possible explanations of the origin of this extended damage layer. Point defects generated during implantation can diffuse from the implanted layer into deeper regions [25]. However, this is in contradiction to the results of Kinomura et al. They excluded diffusion because their iron samples were irradiated at low temperature of

Fig. 3 The depth distribution of the positron lifetime for well-annealed silver samples irradiated with Xe^{26+} energy of 167 MeV with two doses, tagged in the figure. The gray rectangle represents the implanted layer. The hatched region represents the bulk value of the positron lifetime in silver

100 K where vacancies are immobile and despite this they observed an extended damage region beyond the projected range [13]. Another explanation was suggested by Sharkeev and Kozlova [5]. A high level of mechanical stress caused by ion implantation may induce generation of defects not only in the implanted layer but as well beyond it. To confirm one of these explanations we carried out the following experiment: the iron sample was irradiated with Xe^{26+} ions after passing the aluminum 6- μm -thick foil. After irradiation, the sample was annealed at 150 °C during 1 h. At such a temperature, we expect rather recovery of defects and stress relaxation than deeper changes like recrystallization. For significant change in the iron structure, e.g., recrystallization, the temperature of about 400–450 °C is necessary, [26] (see also Ref. [27]). Thus, we expected that in this temperature the thermally activated dislocation motion under internal stresses will occur, resulting in the dislocation density reduction. In Fig. 4 we depicted the results. It can be seen that the defects remained only in the implanted layer. Beyond this layer, the mean positron lifetime reaches the bulk value indicating only residual defects. This experiment excludes diffusion of defects beyond the implanted layer, because this could cause an increase of the mean positron lifetime in this region. Therefore, we accept the point of view that in the extended damage layer the observed defects are generated by the high level of stress induced during implantation. To support this, a simple model was proposed and theoretical calculations of stress, using finite element analysis were performed.

Fig. 4 The depth distribution of the mean positron lifetime for well-annealed iron sample irradiated with Xe^{26+} of 167 MeV after passing the aluminum foil 6.5 μm thick. The corresponding projected range in iron is equal to $R_d = 5.8 \mu\text{m}$. Then, the sample was annealed at 150 °C during 1 h. The gray rectangle represents the implanted layer. The hatched region represents the bulk value of the positron lifetime in iron

4 The model with the use of finite element analysis

Let us consider a simple model of a plate 0.3 mm thick. At the top, a 10- μm -thick layer was chosen, in which the uniform stretching stress is applied, Fig. 5. This layer would correspond to the implanted layer in our experiments presented above. The model was chosen to be 2D plane strain cross section through the plate. Simple beam supports were used as shown in Fig. 5 a. The length was shorter, than actual size of the plate in the experiment, to reduce the size of computations. The 1 mm length was enough to obtain equalized stress field in X -direction through the middle portion of the model.

The calculations were performed for two plates: iron, where Young's modulus is equal to $E=210$ GPa, Poisson's ration, $\nu=0.3$, and yield strength $Y=150$ MPa; and silver with $E=83$ GPa, $\nu=0.37$ and $Y=55$ MPa. Material model with perfect plasticity was set. The part was meshed with a three-noded linear plane strain triangle element CPE3. The mesh was defined in the layer area with an approximate size of 2 μm in the stretched layer and 25 μm at the bottom, Fig. 5a.

The top layer was stretched in one direction, parallel to the plates' plane. This was done by setting isotropic expansion property to the layer, with expansion coefficient only in X -direction. Then, uniform temperature

load was applied to this part. The temperature load was raised until von Mises stresses in the layer reached yielding point. It can be seen in Fig. 5 b that the plate bent slightly. The depth distribution of the stress in the perpendicular cross section (AA) is depicted in Fig. 6 for two metallic plates. The gray rectangle represents the layer where the stretching is added. In this layer, the applied stress is about -150 MPa; however, beyond this layer, the stress increases to the value of about 21 MPa at the depth of 14 μm and then slightly decreases with the gradient of about 0.116 MPa/ μm . One should note that the stress changes the sign and is equal to zero at a certain depth. This is due to the fact that the layer is stretched, and the deeper layers are consequently squeezed. The obtained results, in principle, correspond to the results of calculations performed by Sharkeev and Kozlova (see Fig. 9 in Ref. [5]). However, we focus our interest in the region near the implanted layer, which means the stretched top layer.

To link the results of positron annihilation with the obtained stress distribution shown in Fig. 6, we must return to our former results in Ref. [28, 29]. There was established the coincidence of the von Mises criterion for yield with the onset of the increase of the defect concentration tagged by the increase of annihilation line shape parameter, which also corresponds with the mean positron lifetime. In other words, the increase in the mean positron lifetime results from an increase in the stress above the yield strength. The yield strength is the beginning of plasticity, i.e., generation

Fig. 5 The model of the plate with the stretched layer on the top with displayed mesh (a). Map of the stress in direction X (b) in MPa. Deformation magnified by factor $\times 200$. The cross section AA for determination of the stress distribution across the layer

Fig. 6 The stress distribution calculated for iron (a) and silver (b) according to the model, see text, as a function of the depth from the surface. The gray rectangle represents the region of the stretched layer. The solid lines represent the results when the stress in the stretched layer is about 150 and 85 MPa for iron and silver, respectively. The dashed lines constitute the stress distribution when the stress in the stretched layer was 1 and 0.5 GPa for iron and silver, respectively. The dotted horizontal lines tag the yield strength Y in both metals

of dislocations and other defects. Below this, only elastic deformation takes place which does not damage material and does not produce any defects which could be tagged by the positron annihilation characteristics. This has been demonstrated in numerous studies of various metals, see, e.g. Ref. [27]. Taking this into account, we can conclude that only in the stretched layer (gray rectangle) the increase of the positron lifetime is possible. Beyond this region, only the elastic deformation is possible, because the values of the stress are below 21 MPa, for iron, Fig. 6 a, and 8 MPa, Fig. 6 b for silver. This is much below the yield strength for these metals. Thus, we should observe an increase of the positron lifetime above the bulk value only in the region of the stretched layer, not beyond. These calculations indicate that the presence of the stress close to the yield strength in the implanted layer results in some distribution of the stress in much deeper region as well. However, it should not be reflected by any increase of the mean positron lifetime because the stress in this region is too small to generate defects.

One can suppose that the increase of the stress above the yield strength in the implanted layer may induce the increase of the stress beyond this layer. For showing this, we rejected the plasticity from the calculations, and the initial stress in the calculations was increased up to value of 1 GPa for iron, and 0.5 GPa for silver. The long dashed lines represent the results of such calculations, Fig. 6. It can be noticed that now the stress beyond the stretched layer exceeds the yield strength value, tagged by dotted horizontal line, however, only at a certain depth. For iron, this is between 13.5 and 20 μm, and for silver between 13.5 and 40 μm. Indeed, it was reported that ion implantation can generate static and dynamic stresses to the value of 1 GPa at a high ion dose

[30]. EerNisse and Picraux found the average stress in the He-implanted layer at least one order of magnitude greater than the yield strength [31]. Thus, one can assume that such large stress may be obtained during ion implantation, and this can perfectly explain the experimental results shown in Figs. 2 and 3. The small amount of defects can be distributed up to the depth of 20 μm in case of the iron samples and much deeper in case of silver samples. Silver is a ductile metal; therefore, it can be damaged easier than iron, thus the extended damage layer can extend much deeper.

One can point out also another interesting feature, which supports the proposed model. In Fig. 6, it can be seen that the stress is below the yield strength beyond the stretched layer at a depth in the range for 11.7–13.0 μm. Therefore, at this depth, there should be no defects. Deeper their concentration should increase and then decrease. Thus, in the extended damage layer, a certain maximum should be observed. Indeed, such a maximum seems to be visible in Fig. 1, for the 9.5-μm-thick foil. Nevertheless, this effect is more pronounced in titanium (see Fig. 2 in Ref. [19]) or in silver [18] as well. Maximum was also noticed in the dislocation density in a distance from the entrance surface of copper irradiated with different ions, i.e., Ti of 105 keV, Hf of 100 keV and Zr of 130 keV as it is reported in Ref. [4, 32]. (see also Fig. 10.24 in Ref. [2]). Nevertheless, this maximum is located at the depth of about 10 μm, whereas the projectile range in their experiment was about 0.037 μm for Ti ions and 0.025 μm for Zr ions. Thus, the maximum should be observed at much lower depth. Nevertheless, it is not excluded that stretched layer, which was used in the model, in some cases can be much thicker than the implanted layer, because other processes can be also an origin of the

stress, e.g. a significant increase in temperature, blistering etc.. One should mention a model proposed by Sharkeev et al. [4]. Within their model the stress generated in the implanted layer induces motion of dislocations, which are moving beyond. However, such a model does not explain the maximum in dislocation density in the extended damage layer.

We can conclude that the presented model explains, observed by positrons, the extended damage layer beyond the implanted layer. It can be stated carefully that in this layer during implantation the dynamical stress exceeds the yield strength several times. The observed effect depends on the metal and its mechanical properties. It can be assumed that observed by Sharkeev and Kozlova LRE can be explained by this model. This is supported by the fact that in their experiments the much higher dose of 4.6×10^{17} ions/cm² was applied. This could induce much higher stress in the implanted layer and then beyond, and it can affect the sample at large depths, as reported in Ref. [5].

It is possible to consider whether the obtained positron results can be used to determine the stress in the implanted layer. The thickness of the expanded damage layer appears to be very helpful in this determination because it is correlated with the stress in the implanted layer and the mechanical properties. The applied model predicts linear dependency between the stress inside the stretched layer σ_s and the maximal value of the stress in the extended layer σ_m : $\sigma_m \cong -0.147 \times \sigma_s$. This relationship is valid for iron and silver.

One can point out two causes of the stress in the implanted layer. The thickness of this layer is very small, thus the deposited by swift ions energy can lead to increase of the temperature. The cooling system that cools the irradiated sample may not take away the whole energy deposited by the ions. This generates thermal expansion that induces the bending of the implanted sample as shown in Fig. 5b. Deposition of ions in the sample can cause swelling of the implanted layer which can give the same results as in Fig. 5b. The striking feature is that the value of the stress inside the stretched layer exceeds several times the yield strength. This should lead to the detachment of this layer from the rest of the plate. Nevertheless, as it was pointed out by EerNisse and Picraux, “the ability of the material to support stress larger than yield strength in the plane stress case of a very thin attached layer is well known in the fields of thin-film deposition and oxidation of metals and semiconductors” [31]. One can suppose that the existence of the extended damage layer can affect the properties of products subjected to the implantation process.

5 Conclusions

The implantation of swift Xe²⁶⁺ ions with different energies into iron and silver plates induces generation of the implanted layer with defects such as vacancy clusters and dislocations in case of iron, and dislocations in case of silver, which were detected using conventional positron lifetime spectroscopy with sequenced etching of the irradiated samples. The projectile ranges, calculated from SRIM code for different ion energies, correlate with the thickness of the implanted layer detected using the positron annihilation technique. However, apart from this layer, an extended layer of damage is observed at a depth about twice as large as the depth of the implanted layer. The origin of this layer has been explained as the distribution of the stress, which exceeds in the implanted layer the yield strength. A finite element method using ABAQUS 2017 software has shown that in the extended damage layer the maximum in defect concentration is present, what is experimentally confirmed in the positron lifetime measurements. However, for the ductile silver, the thickness of this layer is much about 40 μm and this is also observed in the measurements. The calculations allow us to estimate the stress in the implanted layer to about 1 GPa in iron and about 0.5 GPa in silver. These results can explain LRE observed by other authors.

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Appendix 1

The implantation profile of positrons, $p(z)$, i.e., number of positrons which are implanted in the layer dz thick at the depth of z , emitted from β^+ radioactive isotopes into matter is usually expressed by the single exponential decay function with only one parameter, i.e., linear absorption coefficient μ . Its reciprocal value is called the positron implantation range. However, after a closer inspection near the source, the profile changes its shape in such a way, that a second exponential function can be noticed, see, e.g., Ref. [33]. Thus, it is the short-range component. The good approximation of the implantation profile as the function of depth from the entrance surface (z) can be expressed as follows:

$$p(z) = \begin{cases} \frac{N}{2} [\exp(-2\mu z) + \exp(-2\mu d)], & 0 < z \leq d, \\ N \exp[-\mu(z + d)], & z > d, \end{cases} \quad (2)$$

where $N = 2N_0[1 + \exp(-2\mu d)]^{-1}$, N_0 is the total number of emitted positrons and $d \cong (2\mu)^{-1}$. In this approximation, the linear absorption coefficient for the short-range component is about twice of the μ value. Such a formula allows us to express the implantation profile using only one experimental parameter, i.e., μ . However, recently, Dubov et al. proposed to use more parameters for description of this profile but still two exponential decay functions were applied [34]. Another formula that takes into account the energy spectrum of emitted positrons can also be found in the Ref. [35].

This indicates that more positrons annihilate near the entrance surface, which would result from a single exponential decay profile. This broadens the application of the etching technique, described above, to the studies of defects located at much shorter range than at the distances $1/\mu$. The linear absorption coefficient for iron is about 389 cm^{-1} , thus in Eq. (2) $d = 12 \text{ }\mu\text{m}$. Taking into account these values, in the $8.5\text{-}\mu\text{m}$ -thick layer of the surface about 37% of positrons are stopped, whereas taking into account the single exponential decay function, only 29% would stop there.

The linear absorption coefficient for the short-range component in the implantation profile can be obtained from the experimental results presented above. During deconvolution of the positron lifetime spectra, one can assume that one component is equal to 109 ps. This corresponds to the bulk value which is present only beyond the implanted and extended damage layers. The intensity

Fig. 7 Intensity of the lifetime component equal to 109 ps obtained from the deconvolution of the positron lifetime spectra measured in the etching experiment for iron sample irradiated with the Xe^{26+} ions. This corresponds to positrons, which annihilate in non-damage region, beyond the implanted and extended damage layer. The solid line represents the best fit of the following function: $I(z) = 100[1 - \exp(-\mu_s(z + c))]$ to the experimental points. The value of the adjustable parameters are equal to $c = 9.0 \pm 0.2 \text{ }\mu\text{m}$ and $\mu_s = 1383 \pm 38 \text{ cm}^{-1}$

of such a component should inform about fraction of positrons which pass these layers and annihilate beyond them. In Fig. 7, such an intensity as the function of depth from the entrance surface is depicted. From this, one can determine the value of the linear absorption coefficient for the short-range component, i.e., $\mu_s = 1383 \pm 38 \text{ cm}^{-1}$, Fig. 7, and it is about three and a half larger than μ value above.

Appendix 2

Simple considerations lead to a relationship which links the positron implantation profile $p(x)$, the local value of the mean positron lifetime at a certain depth x : $\bar{\tau}(x)$ from the entrance surface, and the measured mean positron lifetime $\bar{\tau}_m(z)$ after etching the layer of thickness z ; as the integral equation:

$$\bar{\tau}_m(z) = \int_z^\infty p(x - z) \bar{\tau}(x) dx. \quad (3)$$

Assuming that implantation profile is expressed as the exponential decay function with linear absorption coefficient μ , the solution of this equation is as follows:

$$\bar{\tau}(z) = \bar{\tau}_m(z) - \frac{1}{\mu} \frac{d \bar{\tau}_m(z)}{dz}. \quad (4)$$

Thus, the measured mean positron lifetime reproduces the local mean positron lifetime after its gradient correction. If the thickness of etched layer is small, one should apply the short-range component in the implantation profile and in (4) also μ should be replaced by μ_s . Thus, the correction for the gradient is even smaller.

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