Positron annihilation techniques applied to reactor steels

Vladimír Slugeň, Róbert Hinca and Matúš Stacho
Slovak University of Technology, FEI, Ilkovičová 3, SK-81219 Bratislava, Slovakia
E-mail: Vladimir.Slugen@stuba.sk

Abstract. Different positron annihilation spectroscopy (PAS) techniques were applied in investigations of nuclear reactor steels. Summary of previous applications and evaluation of PAS potential contributions in the area of microstructural study of materials foreseen for fusion and fission technologies are discussed in detail.

1. General view to PAS applications in steels
If we consider about positron annihilation from the general point of view, we could say that the positron annihilation spectroscopy (PAS) is nowadays well recognized as:
- a powerful tool of microstructure investigations of condensed matter,
- a spectroscopic technique for the study of vacancy type defects,
- a technique in material study for registration also very low concentration of defects,
- one of suitable techniques for defects study in the near surface region [1-4].

The utility of positron annihilation studies of reactor pressure vessel steels (RPV) relies on the fact that the characteristics of the annihilation process depend almost entirely on the initial state of the positron-many-electron system. Since energetic positrons are rapidly thermalized after entering condensed matter, if they are not bound to electrons, the characteristics of the annihilation process in most cases depend on the initial state of the many-electrons system where positrons annihilate. The energies, moment, and time of the gamma rays emitted during the annihilation may be measured with high precision using the modern detector systems. Therefore, studying the positron annihilation process characteristics can identify the state of electrons in metallic substances.

The annihilation lifetime can be determined by lifetime measurement. Momentum distribution of the annihilation gamma rays can be measured using the method of angular correlation of the two collinear annihilation gamma rays, and energy distribution using the method of Doppler broadening.

Hence, while the lifetime method yields information regarding the electron density in the region of electron-positron overlap through the determination of positron decay rates from various states in the metal. The method of angular correlation and Doppler broadening (momentum technique) gives information regarding the electron moment in this overlap region through the shape of the angular (Θ) and energy (E) distributions of the annihilating γ-rays. These shapes are usually characterised by one or more "line shape parameters". The most commonly used ones are the normalised peak-counts determined in an interval around 0 (in angular correlation) or $E = 0$ (in Doppler broadening). These measure in various, but qualitatively equal, ways the change of annihilations with electrons of low momentum relative to those with high momentum. These changes in the annihilation characteristics yield information whether the positron annihilates in a defect-free region of the crystal or from a defect site.
2. PAS applications on RPV steels
Since mid 1980-s PAS has been extensively used in the study of RPV steels [5-19]. The positron annihilation technique can give essential information about the deterioration of the mechanical properties of RPV steels during their irradiation, which is known as neutron embrittlement. The lifetime of positrons trapped at radiation-induced vacancies, vacancy-impurity pairs, dislocations, microvoids, etc. is longer than that of free positrons in perfect region of the same material. As a result of the presence of open-volume defects, the average positron lifetime observed in structural materials is found to increase with radiation damage.

The embrittlement of steel is a very complicated process depending on many factors (thermal and radiation treatment, chemical compositions, preparing conditions, ageing, etc.). Properties of the RPV steels and influence of thermal and neutron treatment on these properties are routinely investigated by macroscopic methods such as Charpy V-notch and tensile tests. A number of semi-empirical laws, based on macroscopic data, have been established, but, unfortunately, these laws are not completely consistent with all data and do not provide the desired accuracy. Therefore, many additional test methods [20-22] have been developed to unravel the complex microscopic mechanisms responsible for RPV steel embrittlement.

RPV embrittlement poses one of limiting factors in the lifetime of vessels of today’s nuclear power plant (NPP) This problem is very serious in Eastern (Russian) types of nuclear reactors (VVER). It is due to the narrower gap between the outside surface of the core barrel and the inside surface of the RPV than in Western RPV’s. The neutron flux and consequently neutron fluency on the RPV wall is generally higher for VVER-440 type reactors than for other equivalent types. This influence of neutron flux (even neutrons of energy over 0.5 MeV) on RPV embrittlement is much more pronounced than other contributions, e.g., from a coolant temperature or from the operational pressure in the primary circuit.

According to previous reports [23-25] it seems to be generally accepted that even in the Western types of RPV steels containing more than 0.1 wt. % of Cu, the Cu- and P-rich precipitates play a dominant role in thermal and neutron embrittlement. In the case of Eastern-type RPV steels, comprehensive PAS studies [26-28] have suggested that carbide formation is an observable additional microstructural mechanism.

Positron annihilation can be used to recognize and study the following features and properties of RPV steels.

2.1. Differences in reactor pressure steel according to chemical composition
Comparison of 5 western and 3 Russian RPV steels using PAS LT techniques was published in [18] (see figure 1) and table 1. The increased amount of alloying elements causes larger grain boundaries and nonhomogeneity which result to increased positron lifetime.

| Code | Type of steel | Contents of alloying elements in RPV specimens (wt.%) |
|------|--------------|------------------------------------------------------|
|     |              | C | Si | Mn | Mo | Ni | Cr | Cu | P | S | V | Co | Total |
| EGF | 22NiMoCr37   | 0.22 | 0.23 | 0.88 | 0.51 | 0.84 | 0.39 | 0.080 | 0.006 | 0.004 | 3.160 |
| D4  | A508B Cl.3   | 0.20 | 0.28 | 1.43 | 0.53 | 0.75 | - | 0.055 | 0.008 | 0.008 | 3.261 |
| CH  | 18MND5       | 0.18 | 0.26 | 1.55 | 0.50 | 0.65 | 0.18 | 0.140 | 0.007 | 0.002 | 3.469 |
| JRQ | A533B Cl.    | 0.18 | 0.24 | 1.42 | 0.51 | 0.84 | 0.14 | 0.140 | 0.017 | 0.004 | 3.491 |
| 73W (weld) | Linde 0124  | 0.10 | 0.45 | 1.56 | 0.58 | 0.60 | 0.25 | 0.310 | 0.005 | 0.005 | 3.860 |
| YA  | 15Ch2MFA     | 0.14 | 0.31 | 0.37 | 0.58 | 0.2 | 2.64 | 0.090 | 0.014 | 0.017 | 0.27 | 0.019 | 4.650 |
| YTA (HAZ) | 15Ch2MFA  | 0.14 | 0.24 | 0.40 | 0.72 | 0.14 | 2.93 | 0.110 | 0.013 | 0.017 | 0.31 | 0.011 | 5.031 |
| XTA (HAZ) | 15Ch2NMFA  | 0.18 | 0.24 | 0.52 | 0.62 | 1.26 | 2.22 | 0.080 | 0.010 | 0.013 | 0.08 | 0.008 | 5.231 |
2.2. Differences in the same type of steels according to different thermal treatment

PAS techniques are very sensitive on annealing. Many works were focused on registration of optimal annealing procedures. One of the most appreciate picture in the “positron community” from industrial point of view was PAS confirmation of optimal annealing temperature of 15Kh2MFA RPV steel (see figure 2). Next picture shows the temperature region of annealing and informs that annealing at 500 ºC causes creation of next large amount of small defects connected to hug probably carbide precipitation.

2.3. Irradiation effect

The positron annihilation can register an increase of vacancy type defects due to irradiation [5,12,15], however, the loads during RPV operating lifetime $10^{20}$ n/m$^2$ are too small to register large changes in positron annihilation parameters. Irradiation temperature at level of about 300 ºC causes partial annealing of small (vacancy) defects, so that positron lifetimes stay practically unchanged or decrease after initial period (probably after Cu, P and/or carbides precipitation and connected re-structuralisation). For more information see [29] and figure 3.

Figure 1. Comparison of positron mean lifetimes according to different chemical composition of RPV steels.
2.4. Depth profiling
Interesting depth profile was observed in the near surface region in base as well as weld metal in case of Russian WWER-440 RPV steels. The increased defects concentration after initial period of irradiation is much more observable in the near surface region. It decreases with the depth into the bulk. A plausible explanation is connected to the Cu precipitation and segregation on the grain boundaries.
Figure 4. The 3D presentation of PLEPS results in form of the MLT dependence on neutron fluence in WWER-440 weld (ZK) and base metal (ZM). The first about 140 nm were neglected due to possible surface defects.

2.5. PAS parameters in comparison to hardness
Similar process (Cu, P and C precipitation) as in D can be confirmed also using measurements of WWER-440 RPV steels published in [18]. Figure 5 shows a comparison of hardness and PAS parameters versus Hollomon-Jaffe´s parameter which calculate temperature and time of annealing in one.

2.6. Recognition of nano-size defects
Positron annihilation can recognize also nanosize-defects. In figure 6 the positron lifetime is assigned not only to number of vacancies in defect but also to its diameter. The graph shows two curves obtained using two different enhancement factors: the one from Boronski-Nieminen [30] and the one from Stachowiak-Lach [31]. In the case of ref. [32] we can see a somewhat too slow saturation of the lifetime with the increasing size of the cluster. The reason is the incorrect behaviour of the corresponding enhancement for low electron densities (although the limit for zero electron density is correct). This is avoided in the approach of ref. [31], but the resulting lifetimes are a bit too short for
small vacancy clusters. The positron surface state was not considered in these calculations [33]. These calculations were used also for positron study of ITER first wall materials [34].

2.7. Positron affinity calculations
From the point of view of PAS, coherent and incoherent particles exhibit different behaviour. In the latter case, there are free volumes at the matrix-particle interface and positrons may get trapped there. But this does not apply to a coherent particle, and positrons can be localised inside such a particle. Coherent particles may represent either a potential barrier or a potential well for positrons. An important quantity to evaluate the behaviour of the potential in the vicinity precipitates is the positron affinity. The change of the positron affinity at the matrix-particle interface determines the step of the positron potential.

In order to have some idea whether the positron trapping at precipitates may occur, the positron affinity calculations for Cu and possible precipitates which could exist in CuCrZr and CuAl25 systems were performed [34]. Table 2 presents the obtained results.
The necessary condition for positron trapping at an embedded particle is that the positron affinity of such a particle lies below the positron affinity of the matrix. One can see that this is not fulfilled for any considered particle except for Al₂O₃. In the case of Zr and Cu₂Zr there is no difference in affinities calculated using the GC computational scheme (within the precision of calculations). Thus, the positron trapping cannot be excluded, but it is not very likely. Another aspect is that the particle must have a certain radius (size) in order to hold at least one positron level [34]. When the difference in affinities is small, the necessary radius may become very large. Table 2 also contains calculated positron lifetimes together with their available experimental counterparts.

### Table 2. Summary of affinity ($A_+$) and lifetime ($\tau$) calculations for Cu and possible precipitates. $A_+$ and $\tau$ are in units of eV and ps, respectively. Available experimental results are also available.

| System | $A_+$ (Cu) | $A_+$ (Cr) | $A_+$ (Zr) | $A_+$ (Cu₂Zr) | $A_+$ (α-Cr₂Zr) | $A_+$ (α-Al₂O₃) |
|--------|------------|------------|------------|---------------|----------------|-----------------|
| BN     | -4.8       | -2.8       | -4.5       | -4.6          | -4.2           | -7.3            |
| GC     | -4.1       | -2.3       | -4.1       | -4.1          | -3.7           | -6.8            |
| Exp.   | -4.2       | -2.4       | -4.2       | -2.4          | -4.2           | -6.8            |
| BN     | 104        | 98         | 154        | 114           | 120            | 138             |
| GC     | 118        | 104        | 154        | 126           | 124            | 152             |
| Exp.   | 110        | 120        | 110        | 120           | 120            | 152             |

2.8. Comparison of calculated and measured positron lifetimes

It is not easy to assign the positrons lifetime only to one type of defects. The calculated positron lifetimes for different types of defects in pure iron and different carbides in low alloy Cr-Mo-V steel are shown in table 3.

### Table 3. Calculated or measured positron lifetimes in Fe or in carbides.

| Material               | Positron lifetime (ps) | Reference |
|------------------------|------------------------|-----------|
| Fe-bulk                | 110                    | [35]      |
| Fe-dislocations        | 165                    | [36]      |
| Fe-monovacancy         | 175                    | [35]      |
| Fe-divacancy           | 197                    | [36]      |
| Fe-3 vacancy cluster   | 232                    | [36]      |
| Fe-4 vacancy cluster   | 262                    | [36]      |
| Fe-6 vacancy cluster   | 304                    | [36]      |
| VC                     | 99                     | [37]      |
| $V_{0.86}Cr_{0.09}Mo_{0.04}Fe_{0.01}C$ | 105               | [37]      |
| $Mo_2C$                | 112                    | [37]      |
| $Mo_{1.4}Cr_{0.6}C$    | 116                    | [37]      |
| Cr₅C₃                  | 107                    | [37]      |
| Cr₂₃C₆                 | 112                    | [37]      |
| Mn₂₆C₆                 | 99                     | [37]      |
| Fe₃C                   | 101                    | [37]      |

However, due to FWHM on the level of 200 ps it is difficult to distinguish between all types of defects. The common approach in this area is grouping defects (mostly into 2 groups). One group comprise small defects like dislocations and mono-, di- vacancies together with bulk value and the second lifetime is mostly assigned to vacancies clusters. There were several trials to distinguish between edge and screw dislocations, which are characterised by lifetimes of 145 (screw) and 165 ps.
According to this work positrons are trapped mostly at dislocation lines (65-85 % of positrons in Russian RPV steel 15Kh2MFA had lifetime on the level 150 ps).

2.9. Calculation of defects concentration from PAS results
Of particular interest is the concentration of defects as a function of irradiation dose and thermal treatment. Generally, this effect can be obtained from positron studies only if the bulk lifetime can be resolved from the shortest lifetime, attributed to annihilation from defects. In the present case of saturation trapping this was impossible. Therefore, from the individual lifetime spectra we can only conclude a total trapping rate \( \kappa \) larger than about \( 10^{10} \text{s}^{-1} \).

However, from the variation of the mean lifetime \( \tau_{av} \) as a function of positron implantation energy, we can estimate \( \kappa \) even in the case of saturation trapping. The problem was fully analyzed in [39,40]. Based on this theory the total defect concentration can be estimated as:

\[
C_d = \frac{\kappa}{\kappa_{\text{spec}}} = \mu \frac{I_2}{I_1} \left( \frac{1}{\tau_b} - \frac{1}{\tau_d} \right)
\]

where
\[
i_1, i_2 \text{ are intensities of positron lifetimes},
\]
\[
\tau_b \text{ – positron lifetime in bulk}
\]
\[
\tau_d \text{ – positron lifetime in defect}
\]

In figure 7 and 8 results for the evaluation of \( \kappa \) (trapping rate) as obtained by the procedure described, are presented. For the specific trapping rate \( \kappa_{\text{spec}} \) the plausible value \( 10^{15} \text{s}^{-1} \) [41] has been assumed. Because of slight surface oxidation, this evaluation of \( \kappa \) results in a systematic underestimate. A lower limit of 10ns\(^{-1} \) for \( \kappa \) may be derived within the framework of the STM since we have observed saturation trapping at defects.

According to the results from our measurements performed on different irradiated RPV-steels, the total trapping rate \( \kappa \) in ns\(^{-1} \) as well as the total defect concentration \( c_d \) (the same values but in ppm) increases slightly for both base and weld materials as a function of the irradiation dose (see figure 8). The weld material (Sv10KhMFT) seems to be less sensitive to the changes caused by neutron-irradiation or by post-irradiation heat treatment than the base material (15Kh2MFA) (see figure 7). Nevertheless, the differences in the positron trapping rate \( \kappa \) are not too large. It seems reasonable to relate the observed trapping rates with the ones which have been derived for trapping into precipitated carbides from electron microscopic images [15]. Accordingly, in the 15Kh2MFA the trapping rate into chromium carbides (Cr\(_7\)C\(_3\), Cr\(_23\)C\(_6\)) is predicted as \( \kappa_{\text{Cr}}=1.8 \times 10^8 \text{s}^{-1} \) and into vanadium carbides as \( \kappa_{\text{VC}}=2.2 \times 10^{10} \text{s}^{-1} \) [15]. Thus precipitated vanadium carbide could indeed account for the observed trapping rates. But on the other hand, as shown by calculations [35], positrons experience a repulsive potential from carbides embedded in an iron matrix. Thus only the defects at the iron-carbide interface could provide an acceptable trapping site for positrons.

The total trapping rate \( \kappa \) and the defect concentration \( c_d \) are stable or increase slightly for 15Kh2MFA steel as a function of Hollomon-Jaffe’s parameter which calculates the influence of annealing temperature \( T \) [K] and time \( t \) [s] together [5,42]. Between 15.5 to 16.5 (corresponding to the temperature region 400 to 450°C) the defect concentration increases. On the other hand (figure 8), in the same range, there is a marked decrease in the lifetime of defects. A simple explanation could be the dissolution of precipitates and defect clusters which would reduce the average size of the defects, and, by the same process, would increase the concentration of the defects.
2.10. Applications of positron microscope
In the positron annihilation spectroscopy, some new lifetime techniques like positron microscope [42] seems to be very promising in defect study. Figure 9 shows a comparison of defect image by cathodoluminescence (CL) and SEM versus pictures obtained from positron microscope in Munich (University of Bundeswehr). The most important problem is actually the relative low positron beam intensity. It can be significantly improved after moving this unique microscope to FRM II reactor in Garching.

2.11. Possible studies of near-surface region
Positron microscope enables very effective material study in the near surface region. For the enhancement of this a special approach was developed in PAS group in Munich based on the cutting of specimen in very small angle. Detailed explanation can be found in [43,44].
2.12. Evaluation of high implanted (irradiated) regions in nuclear materials

Defects in the form of vacancies (loops, voids, etc.) in selected Cu-alloys (99% of Cu) foreseen for applications in thermonuclear fusion technologies were studied using the improved Pulsed Low Energy Positron System (PLEPS), developed at the Institut für Nukleare Festkörperphysik, Universität der Bundeswehr, München [45]. This system enables the study the micro-structural changes in the region from 20 to about 500 nm (depth profiling).

For the simulation of the radiation damage due to neutrons, the ion implantation of protons has been applied. The protons were chosen having approximately the same mass as a neutron. We
supposed that the ballistic influence of protons at the primary-knock-atoms (PKA) production could simulate the ballistic influence of neutrons at Cu-alloys in the fusion reactor ITER. The selected specimens were irradiated in the Ion Beam Laboratory of FEI STU Bratislava [46]. The energy of the implantation was EH=2×95 keV for the molecular H$_2^+$ ion beam (implantation dose up to 1.3×10$^{19}$ ions/cm$^2$ (1.1 C/cm$^2$)). For the protons with the energy of 95 keV, the range of 480 nm was found by TRIM calculations. It optimal corresponds to the scope of the PLEPS equipment. Results enable not only to compare the implanted damage profile to positron lifetimes, but also to study the annealing behaviour of studied materials. By higher implantation doses the blistering effect can be studied, too [47].

![Figure 11.](image)

**Figure 11.** Annealing behavior of H$_2^+$ ions implanted 99CuZrCr alloys [48].

2.13. Evaluation of role of Cr in Fe-Cr alloys

Depth profiling of defects down to one micrometer below the surface has been performed on an untreated Fe11.62%Cr sample and three Fe11.62%Cr samples with different damage levels using positron implantation energies between 1 keV and 18 keV. The depth profiles were measured with the Pulsed Low Energy Positron System (PLEPS) [49] at the high intensity positron source NEPOMUC [50] at the research reactor FRMII at TUM.

To obtain cascade collisions in the microstructure of the studied materials without neutron activation, accelerated helium ions have been used. The depth profile of collision events can be seen in figure 12. The measured positron lifetimes spectra were analyzed using the LT 9.0 program [51] and a modified version of PosFIT [52]. The differences of these two analyses were negligible. All the lifetime spectra could be decomposed into three lifetime components with variances close to one.

Figure 13a shows the positron mean lifetime as a function of helium implantation dose and mean positron implantation depth. The positron mean lifetime (MLT) is increasing with the implantation dose, thus indicating the creation of defects due to implantation. The increase of the MLT close to the surface (< 200nm below the surface) is probably due to positrons annihilating in surface oxide layer. At higher depths the course of the MLT depth profile corresponds to the expected zone of maximum damage. In the zone of maximum damage in the implanted specimens two different defect lifetimes have been observed. The shorter lifetime between 240 ps and 300 ps could be assigned to small vacancy clusters < 6 vacancies or larger clusters filled with helium. The longer lifetime between 400 ps and 500 ps corresponds to annihilation in large voids (> 1 nm). The intensity of this longer
component \( I_3 \) increases dramatically with the helium implantation dose as can be seen from figure 13b. The course of the \( I_3 \) depth profile again corresponds to the expected zone of maximum damage.

**Figure 12.** Depth profile of the helium implantation, \( E=250\text{keV}+100\text{keV} \) (SRIM simulation of \( 10^6 \) ions).

**Figure 13.** a) Positron mean lifetime in different treated Fe11.62%Cr alloy and b) intensity of annihilation in large defects (voids) measured in Fe11.62%Cr alloy.

### 3. Conclusions
In the last years several original irradiated RPV steel samples were investigated by PAS technique with the aim to observe some micro structural changes due to thermal and neutron treatment resulting from operating conditions in NPP. This investigation is aimed mainly at increasing the safety margin against the RPV brittle fracture. The only possibility to reduce this irradiation embrittlement is thermal annealing. Non-destructive test methods like PAS in combination with proved destructive test methods can contribute to a progress in optimisation of temperature-time regime for post-irradiation thermal treatment. This approach is especially useful in the case of highly radioactive materials and also in those cases where only small amounts of irradiated materials are available. Of course, in the long –
term point of view, PAS techniques can be effectively used in the testing and selection process of optimal reactor steels foreseen for Generation IV reactors and thermonuclear fusion facilities.

Acknowledgement
This study was supported by VEGA 1/3188/06 and 7RP-Euratom/CU.

References
[1] Hautojärvi P 1979 Positrons in Solids (Berlin: Springer-Verlag)
[2] Krause-Rehberg R 1997 Positrons in semiconductors (Berlin: Springer-Verlag)
[3] Dauwe C, Dorikens M, Dorikens-Vanpraet L and Segers D 1974 Appl. Phys. 5 117
[4] Sperr P and Kögel G 1997 Mater. Sci. Forum 255-257 109
[5] Brauer G, Liszay F, Molnar B and Krause R 1991 Nuclear Engineering and Design 127 47
[6] Pareja R, De Diego N, De La Cruz R M and Del Rio J 1993 Nucl. Technol. 104 52
[7] Lopes Gil C, De Lima A P, Ayres De Campos N J, Fernandez J V, Kögel G, Sperr P, Trifshhäuser W and Pachur D 1989 J. Nucl. Mater 161 1
[8] Valo M, Krause R, Saarinen K, Hautojärvi P and Hawthorne R 1992 Effects of Radiation on Materials: 15th International Symposium, ASTM STP 1125 ed R E Stoller, A S Kumar and S G Gelles (Philadelphia: ASTM) p 172
[9] Slugen V, Hašičk J, Gröne R, Bartik P, Zeman A, Kögel G, Sperr P and Trifshhäuser W 2001 Mat. Sci. Forum 363-365 47
[10] Nagai Y, Tang Z and Hasegawa M 2000 Radiation Physics and Chemistry 58 737
[11] Brauer G, Puska M, Sob M and Korhonen M 1995 Nucl. Eng. Desg. 158 149
[12] Slugen V 2005 Nucl. Eng. Desg. 235 1961
[13] Hartley J H, Howell R H, Asoka-Kumar P, Sterne P A, Akers D and Denison A 1999 Appl. Surf. Sci. 149 204
[14] Becvar F, Cizek J, Lestak L, Novotny I, Prochazka I and Sebesta F 2000 Nucl. Instr. Meth. A 443 557
[15] Cizek J, Prochazka I, Kocik J and Keilova E 2000 Phys. Stat. Sol. (a) 178 651
[16] Van Hoorebeke L, Fabry A, van Walle E, Van de Velde J, Segers D and Dorikens-Vanpraet L 1996 Nucl. Instr. Meth. A 371 566
[17] Ghazi-Wakili K, Zimmermann U, Brunner J, Tipping P, Waeber W B and Heinrich F 1987 Phys. Stat. Sol. (a) 102 153
[18] Slugen V, Segers D, De Bakker P M A, DeGrave E, Magula V, Van Hoecke T and Van Waeyenberge B 1999 J. Nucl. Mater. 274 273
[19] Goland A N 1976 Proceedings of the International School of Physics “Enrico Fermi” ed G Carlioti (Amsterdam: North-Holland) p 110
[20] Phythian W and English C A 1993 J. Nucl. Mater. 205 162
[21] Ammaev A D, Dragunov Yu G, Kryukov A M, Lebedev L M and Sokolov M A 1986 IAEA Specialists Meeting Proceedings, Plzen
[22] De Bakker P, Slugen V, De Grave E, Van Walle E and Fabry A 1997 Hyperfine Interaction 110 11
[23] Koutsky J and Kocik J 1994 Radiation Damage of Structural Materials (Elsevier: Amsterdam)
[24] Törrönen K 1979 Technical Research Centre of Finland, Materials and Processing Technology, publication 22, Espoo
[25] Ghoneim M M and Hammad F H 1997 Int. J. Pres. Ves. Piping. 74 189
[26] Davies M 1999 Int. J. Pres. Ves. Piping. 76 163
[27] Becvar F, Jirasková Y, Keilová E, Kocik J, Lestak L, Prochazka I, Sedlak B and Sob M 1992 Mat. Sci. Forum 105-110 901
[28] Brauer G 1995 Mat. Sci. Forum 175-178 303
[29] Slugen V, Zeman A, Lipka J and Debarberis L 2004 NTD &E International 37 651
[30] Boronski E and Nieminen R M 1986 Phys. Rev. B 34 3820
[31] Stachowiak H and Lach J 1993 Phys. Rev. B 48 9828
[32] Dai G H, Moser P and Van Duysen J C 1992 Mater. Sci. Forum 105-110 941
[33] Puska M J and Nieminen R M 1994 Rev. Mod. Phys. 66 841
[34] Slugeň V, Kuriplach J, Ballo P, Domonkoš P, Kögél G, Sperr P, Egger W, Triftshäuser W, Domanková V M, Kováč P, Vávra I, Stanček S, Petriska M and Zeman A 2004 Fusion Engineering and Design 70 141
[35] Puska M J, Lanki P, and Nieminen R M 1989 J. Phys.: Condens. Matter 1 6081
[36] Hahtojärvi P, Pöllönen L, Vehanen A and Yli-Kauppila J 1983 J. Nucl. Mater. 114 250
[37] Vehanen A, Hahtojärvi P, Johansson J, Yli-Kauppila J and Moser P 1982 Phys. Rev. B 25 762
[38] Kocik J, Keilova E, Cizek J and Prochazka I 2003 J. Nucl. Mater. 303 52
[39] Kögel G 1996 Appl. Phys. A 63 227
[40] Slugeň V, Kögel G, Sperr P and Triftshäuser W 2002 Appl. Surf. Sci. 194 (1-4) 150
[41] Brandt W 1967 Positron Annihilation, ed A T Stewart and L O Roelling (New York: Academic Press) p 155
[42] Magula V and Janovec J 1994 Ironmaking and Steelmaking 21 64
[43] David A, Kögel G, Sperr P and Triftshäuser W 2001 Phys. Rev. Lett. 87 067402
[44] Krause-Rehberg R 2003 Material Research in Positron Annihilation (oral presentation at Delft University)
[45] Bauer-Kugelmann W, Sperr P, Kögel G and Triftshäuser W 2001 Mater. Sci. Forum 363-365 529
[46] Kovac P, Bortniansky A, Kloopenkov M and Pavlovets M 2001 Proc. Int. Conf. on Applied Charged Particle Accelerators in Medicine and Industry ACCELERATORS 2001 (Moscow: CNIIatominform) p 430
[47] Constantinescu B and Sarbu C 2000 Fusion Engineering and Design 49-50 171
[48] Slugeň V, Kuriplach J, Ballo P and Domonkoš P 2004 Nuclear Fusion 44 93-97
[49] Egger W, Sperr P, Kögel G and Dollinger G 2007 Phys. Stat. Sol. (c) 4 3969
[50] Hugenschmidt C, Kögel G, Repper R, Schreckeinebach K, Sperr P, Straßer B and Triftshäuser W 2004 Nucl. Instr. Meth. B 211 160
[51] Kansy J 1996 Nucl. Instr. Meth. A 374 235
[52] Kirkegaard P and Eldrup M 1972 Comput. Phys. Commun. 3 240