ELSEVIER

Journal of Nuclear Materials 279 (2000) 301–307

journal of nuclear materials

www.elsevier.nl/locate/jnucmat

# The initial transient of the irradiation growth in a zirconium alloy

A.M. Fortis *, H.C. González

*Laboratorio Daño por Radiación, Departamento Materiales, Comisión Nacional de Energía Atómica, Av. del Libertador 8250, 1429 Buenos Aires, Argentina*

Received 12 August 1999; accepted 7 December 1999

**Abstract**

The onset of the deformation produced by neutron irradiation on a Zr-1.5 at.% Al alloy doped with 0.12 at.% $^{235}U$ is reported. The alloy was subjected to two different thermomechanical treatments: recrystallization (RX) and cold work and stress-relief (CWSR). The measurements were performed in situ in the research reactor RA1 at temperatures close to 350 and at 77 K. Fissioning of the uranium isotope was used as a way of increasing the neutron damage. The growth curves at 350 K show complete coincidence for both thermal treatments up to an equivalent dose of $1–2 \times 10^{23}$ n $m^{-2}$, where they begin to diverge. This result and the behaviour of CWSR at 77 K indicates that the clusters of defects, originated in the displacements cascades, are the main items responsible for the initial growth transient, irrespective of the previous thermal treatment. The equivalent dose at which the long range linear behavior of the CWSR zirconium alloy begins is estimated. 

*PACS:* 61.82.Bg; -61.80.Hg

Irradiation growth is a very well-known phenomenon in the field of materials technology for nuclear reactors [1]. Certain anisotropic materials suffer deformations without or with little volume change when subjected to a neutron flux (especially to fast neutrons), even without external stresses applied on them. Growth has been observed in several hcp materials, particularly in metals like zirconium and its alloys (Zircaloy and Zr–Nb) [2]. Phenomenological studies reveal that irradiation growth depends on temperature, percentage of cold work, crystallographic structure developed, grain size, dislocation density, and obviously, on the irradiation type, intensity and energy [3].

Growth is generally attributed to segregation of vacancies and interstitials to different sinks, particularly to those created by irradiation, which are anisotropically distributed. Collapse of interstitials onto certain crystalline planes produces lattice expansion. As for vacancies, it is generally accepted that they migrate towards sinks in a complex manner. Their evolution with irradiation and their general influence on deformation are not fully understood yet.

In single phase polycrystalline Zr and Zr alloys, growth takes place in the rolling direction, where the highest concentration of $(11\bar{2}0)$ poles appears. An opposite behaviour, i.e., shortening in this direction, has been reported in the case of alloys with more than one phase [4]. The existence of prismatic dislocation loops, with Burgers vector $\langle 11\bar{2}0\rangle$ has been demonstrated [5]. When temperature is increased, they increase in size, decrease in number and appear as a mixture of interstitial loops and vacancy loops, with predominance of the later [6]. Their characterization is difficult at temperatures below 570 K.

A large number of papers have been devoted to the study of materials behaviour in different metallurgical states. The influence of the percentage of cold work, recrystallization texture, grain size, dislocation density, etc. has been analysed [7–9]. Polycrystalline zirconium alloys, either recrystallized (RX) or cold worked and stress-relieved (CWSR), constitute several structural

* Corresponding author. Tel.: +54-11 4754 7352; fax: +54-11 4754 7362.

*E-mail address:* fortis@cnea.gov.ar (A.M. Fortis).

0022-3115/00/$ - see front matter 
PII: S0022-3115(99)00291-3

components of reactors. Generally speaking, growth occurs in the longitudinal direction in both types of alloys. The main difference between CWSR and RX alloys is the growth rate, which is much higher in the former and increases with cold working, without any apparent tendency to saturation at high doses. A linear behavior has been observed for neutron doses higher than $2 \times 10^{24}$ n m$^{-2}$, with $E > 1$ MeV [10]. Recrystallized alloys, on the contrary, reach a certain degree of saturation at intermediate doses or show a decrease of growth rate; at high doses sudden increases of growth rate can also occur [11,12].

In order to correlate a given characteristic behaviour of zirconium and its alloys with the thermal treatment, models have been formulated where the defects, either previously existent or created by irradiation, have a decisive influence. Fitting of these models requires careful experiments, especially with regard to the samples composition and to the neutron flux and spectrum at which they are tested, which must have similar characteristics [13–16].

In a previous work [17] it was shown that the growth kinetics can be measured during irradiation, by means of indirect determinations. Doping with a fissionable material was employed to simulate a high neutron dose and bimetal pairs were used to amplify the measured magnitude and decrease the error. In this work, two specimens with identical composition and different thermomechanical treatment were used to study the initiation of growth deformation and the point where their behaviour begins to be different. One of the samples was tested using the bimetal pair technique; for the second one, growth was measured directly, taking advantage of the sensitivity achieved with the measurement equipment. Both specimens were irradiated in the research reactor RA1, under the same neutron spectrum.

## 2. Experimental method

Postirradiation experiments are reported in the literature where a large difference in growth during the first stages of irradiation was observed in specimens with equal characteristics. Defects reordering due to thermal vibration, occurring during the time elapsed between the irradiation experiment and the deformation measurement is apparently responsible for the randomness of the results, which depends on that time and on the sample temperature. To avoid this problem and with the purpose of obtaining the 'true' growth, measurements in situ, that is, simultaneous with the irradiation, were performed.

The method previously developed consists in the production of damage by means of fission fragments generated within the material. To this purpose, zirconium is doped with a fissionable isotope, in this case $^{235}$U, and irradiated in the core of a research reactor with the standard spectrum of a light water reactor, where the ratio of fast to thermal flux is exactly known. Thermal neutrons produce fissioning of the fissile isotope with a cross-section of hundreds of barns. Each fission fragment, after losing most of its energy through coulomb interactions, has enough energy (about 10% of the initial energy $\approx 7$ MeV on the average) to produce a large number of displacements. This process magnifies the damage provoked by the fast neutrons, i.e., increases the number of produced defects to a level equivalent to a much larger neutron dose. The amplification calculated in the present work is of 20 times (see Section 2.1).

The zirconium alloys were fabricated by electron beam melting in high vacuum. The starting material was high purity Zr and two Al–U alloys: Al $-^{235}$ U and Al $-^{238}$ U, both with uranium contents close to 90 w%. The alloys were subjected to successive fusion steps until most of the aluminum was eliminated; their final compositions are given in Table 1. Gas concentrations were similar to those found in commercial nuclear grade alloys. Strips of $120 \times 3 \times 1$ and $110 \times 3 \times 1$ mm$^3$ were fabricated by rolling and intermediate thermal treatments. The final cold work was 50%.

The longitudinal deformation during irradiation was directly measured in a group of $^{235}$U doped samples stress relieved for 2 h at 670 K in an Ar atmosphere. The texture factor in the rolling direction was $f \sim 0.1$.

On the other hand, bimetal pairs were fabricated by welding at both ends strips of each type of alloy. The pair bends because the strip doped with the fissionable isotope $^{235}$U lengthens due to the action of the thermal neutrons significantly more than the one doped with $^{238}$U, leading to an amplification of the deformation. The amount of bending deformation can be simply related to the differential longitudinal growth of the specimen [17–19]. Moreover, all the effects of length change due to thermal expansion compensate since both components of the pair have received the same thermomechanical treatment and contain the same level of impurities. For this group of samples complete recrystallization was attained by annealing for 2 h at 983 K. Consequently, the resulting texture was more intense than in the former group.

Table 1
Composition of alloys (at.%)

| ZrAl–$^{235}$U | ZrAl–$^{238}$U |
|---|---|
| $U_{total}$: 0, 13 | $U_{total}$: 0, 11 |
| $^{235}$U: 0, 12 | $^{238}$U: 0, 11 |
| Al: 1, 5 | Al: 1, 5 |
| Zr: 98, 25 | Zr: 98, 28 |
| Fe: <300 ppm | Fe: <300 ppm |
| O: <700 ppm | O: <700 ppm |

Two devices, one for each type of specimen, were designed to measure the deformation during irradiation. They were placed in an irradiation loop specially built to reach the core of the experimental reactor RA1, to receive a neutron flux as high as possible and to maintain the samples at a constant temperature during the experiments. Temperatures up to about 750 K can be reached [20] using a low inertia furnace installed within the irradiation loop. Measurements at liquid nitrogen temperature were also performed in a cryostat subjected to the same neutron spectrum [21].

Each sample is supported by a device that leaves the upper end of the sample free to move. The nucleus of a radiation resistant, linear variable displacement transducer (LVDT) is fixed at that end. The holders for both types of samples are made of Zircaloy with the same thermal treatment as the corresponding sample so that the signal variations due to differences in the expansion coefficients are minimized, compensating the dimensional variations provoked by the fast neutrons. In the case of the bimetal pairs, the main restriction is the small room available to place the deformation detector within the facility (Fig. 1(a)). In the case of the direct determinations, a greater versatility exists but a higher precision is required in the measurements (Fig. 1(b)).

The temperature variations obtained during the experiments were in no case superior to 1 K and the thermal gradients were minimized.

### 2.1. Calculation of damage acceleration

The first attempts to measure the number of atomic displacements occurring during irradiation have been made by Kinchin and Pease [22], and Snyder and Neufeld [23]. The modified version of the model of Kinchin–Pease, known as NRT model [24] is generally accepted to give the number of atomic displacements in irradiated materials as a function of the primary knock-on atom (PKA) energy. In this manner, experiments performed with different irradiation sources can be compared.

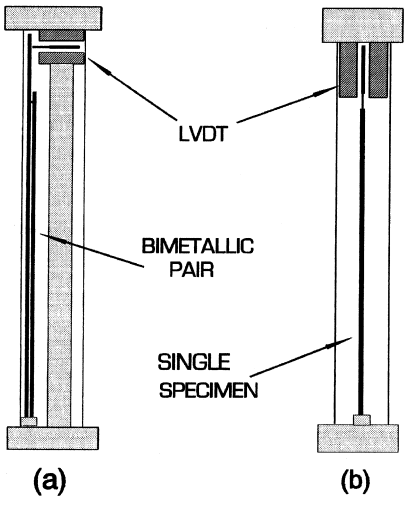


Fig. 1. Devices to measure deformation. (a) Indirect measurement by means of the curvature of a bimetal pair. (b) Direct measurement of the strip elongation.

According to the NRT model, the number of displaced atoms, $\nu(T)$, generated by a PKA of energy $T$ is

$$\nu(T) = \frac{\kappa(T - T_e)}{2E_d} = \frac{\kappa T_D}{2E_d}, \quad T_D > \frac{2E_d}{\kappa}, \tag{1}$$

where $\kappa$ is the displacement efficiency, $T_e$ the energy lost in electronic excitations, $T_D$ the damage energy and $E_d$ is the displacement energy. Calculations of binary collisions indicate that the displacement efficiency is approximately constant ($\kappa = 0.8$) for primary energies higher than $2E_d$ and lower than 100 KeV. Integration of the function $\nu(T)$, over the whole displacement energy spectrum and time, yields the atomic concentration of displacements (dpa, displacements per atom).

The damage energy $T_D$ obtained from the theory of atomic collisions, known as LSS theory [25], can be expressed as

$$T_D = \frac{E}{1 + k_L\, g(\varepsilon)}. \tag{2}$$

$g(\varepsilon)$ can be approximated by [26]

$$g(\varepsilon) = \varepsilon + 0.40244\varepsilon^{3/4} + 3.4008\varepsilon^{1/6}, \tag{3}$$

where

$$\varepsilon = \frac{0.0368 M_2 E}{Z_1 Z_2 [Z_1^{2/3} + Z_2^{2/3}]^{1/2} (M_1 + M_2)} \tag{4}$$

and

$$\kappa_L = \frac{0.0795 (M_1 + M_2)^{3/2} Z_1^{2/3} Z_2^{1/2}}{[Z_1^{2/3} + Z_2^{2/3}]^{3/4} M_1^{3/2} M_2^{1/2}}. \tag{5}$$

In these expressions $M_1$ and $M_2$ are the atomic masses of the projectile and the target, $Z_1$ and $Z_2$ are the respective atomic numbers and $E$ is the energy of the projectile.

Calculations derived from these expressions, based on the energy involved in atomic displacements, do not really permit to establish the number of displacements surviving the annihilation processes (neighbouring interstitial–vacancy pairs, crowdions, superposition of cascades, etc. i.e., all the phenomena subsequent to the collisional phase). Simulation studies and also experimental data obtained at 4 K show that only a minor fraction of the defects predicted by the NRT method survive [27]. All the thermally activated processes, not considered here, must be included to obtain a better description. They considerably reduce the damage, leading to a quantity and distribution of defects quite

different to that obtained with the approximation given above. However, as regards the correlation of the radiation effects in two similar spectra of primary atoms, the number calculated with the NRT method constitutes an adequate parameter for comparative studies.

### 2.2. Amplification of the damage produced by the fission fragments

In order to determine the damage amplification obtained when the material is doped with a fissionable isotope, $^{235}$U, the ratio between the total number of atomic displacements produced by fission fragments and by fast neutrons ($E > 1$ MeV) can be estimated.

#### 2.2.1. Number of atomic displacements produced by fast neutrons

The maximum energy that a fast neutron of energy $E_n = 1$ MeV can transfer to a zirconium atom in an elastic collision is

$$T_{max}(E_n) = 4\frac{mM}{m+M}E_n \approx 43 \text{ keV}, \tag{6}$$

where $m$ is the neutron mass and $M$ is that of a Zr atom.

The numerical approximation of the universal function $g(\varepsilon)$ [26,28,29] given by Eq. (3) yields

$$g(\varepsilon) \cong 2.4 \tag{7}$$

and the damage energy, i.e., the effective energy capable of producing displacements, given by Eqs. (2) and (5), becomes

$$T_D = \frac{E}{1 + k_L g(\varepsilon)} \approx 31 \text{ keV}, \tag{8}$$

where it was assumed $E \equiv T_{max}(E_n)$.

Finally, from the NRT expression (Eq. (1)) the number of displacements, $\nu$, produced by a zirconium atom that has acquired the maximum possible energy of $\approx 43$ keV from a neutron of 1 MeV, is obtained.

$$\nu = \frac{0.8T_D}{2E_d} \cong 500. \tag{9}$$

In this expression the value 25 eV was used for the threshold displacement energy, $E_d$. The value assigned to $E_d$ is not relevant unless displacements in different elements are compared or an absolute theoretical relationship between displacements and a macroscopic property of the material is to be established.

#### 2.2.2. Number of displaced atoms produced by the fission fragments

To calculate the number of displacements in zirconium produced in situ by the fission fragments, the generalized expressions of $\varepsilon$ (Eq. (4)) and $k_L$ (Eq. (5)) are used. The subscripts 1 and 2 indicate in this case the fission fragment and the atoms in the zirconium lattice, respectively.

During $^{235}$U fission more than 200 different isotopes are produced. If the more abundant isotopes produced by fission of $^{235}$U are considered, the relations given in Eqs. (2)–(5) give the damage energy, $T_D$, and the number of displacements, $\nu$, produced by each fission fragment. An average value of $10^5$ displacements per fission event is obtained for zirconium.

#### 2.2.3. Calculation of the production rate of displaced atoms

The total number of displaced atoms per unit time and unit volume can be approximated by

$$\frac{dN_d}{dt} = \phi N_0 \sigma \nu, \tag{10}$$

where $\phi$ is the flux of incident radiation particles, $N_0$ the atomic density of the target material, $\sigma$ the cross-section for the interaction event and $\nu$ is the number of displacements per event.

The flux of thermal neutrons in the reactor RA1 is $5 \times 10^{15}$ n m$^{-2}$ s$^{-1}$, measured using the of $^{59}$Co activation technique; a value of 1.3 is estimated for the fast to thermal flux ratio, with an error of 10%. In these conditions, taking $N_0 = 4.29 \times 10^{28}$ at m$^{-3}$ for zirconium, $\nu = 500$ according to Eq. (9) and finally, considering $\sigma$ equal, on average, to the elastic effective cross section for neutrons of 1 MeV in zirconium, $\sigma_e = 4.5$ barns [30], the production rate of atomic displacements due to fast neutrons per unit volume (Eq. (10)), is $dN_d/dt = 6.3 \times 10^{19}$ m$^{-3}$ s$^{-1}$. If the full neutron spectrum were taken into account and if the average instead of the maximum transferred energy were considered, the number of displaced atoms per primary atom would become lower than 500. A value close to 350 is usually assigned for light water reactors. With these considerations the above expression would result in a lower value.

Finally, considering the thermal flux in the irradiation site, the atomic concentration of $^{235}$U, $\sigma = \sigma_f = 583$ barns and $10^5$ displacements per event, the production rate of displaced atoms due to the fission fragments generated in the doped branch is, by Eq. (10), $dN_d/dt = 1.2 \times 10^{21}$ m$^{-3}$ s$^{-1}$.

In these conditions, the ratio between the damage production rate due to fission fragments and that due to fast neutrons in the irradiated samples, that is, the amplification achieved with a dopant concentration of 0.12 at.% $^{235}$U, is $\approx 20$.

## 3. Results and discussion

An RX bimetal pair was irradiated during 800 h at 365 K. Two CWSR strips were irradiated at 340 and 77

Table 2
Characteristics of the specimens and irradiation conditions

| Thermal treatment | Type of specimen | Irradiation time (h) | Irradiation temperature (K) |
|---|---|---|---|
| RX | Bimetallic pair | 800 | 365 |
| CWSR | Strip | 2000 | 340 |
| CWSR | Strip | 600 | 77 |

K during 2000 and 600 h, respectively, Table 2. All the components doped with $^{235}U$ belong to the same batch, that is, have the same composition. The temperatures reached result from the contributions of the furnace and of γ-heating and vary from one case to another due to the different masses of the samples and devices.

The doses reported in the present work were calculated by the method described in Section 2.2; they were simply obtained by multiplying the fast neutron dose in the specimen's location by a factor of 20.

The results corresponding to 365 K (RX bimetallic pair) and 340 (cold-worked strip), are shown in Fig. 2. A complete coincidence in deformation is observed at the beginning of the irradiation. Then, both curves depart; one of them grows linearly, as is expected for the cold worked and stress-relieved material; the other one, corresponding to the recrystallized material, saturates or grows at very low rate. Deformation of the bimetal pair might be spected to produce stresses which eventually could arrest the growth by irradiation creep. However the measured bending produces a maximum stress of $\approx \sigma_f/30$, being $\sigma_f$ the yield stress for Zr under similar conditions of texture and temperature. Thus, the contribution of irradiation creep is considered negligible in the present case.

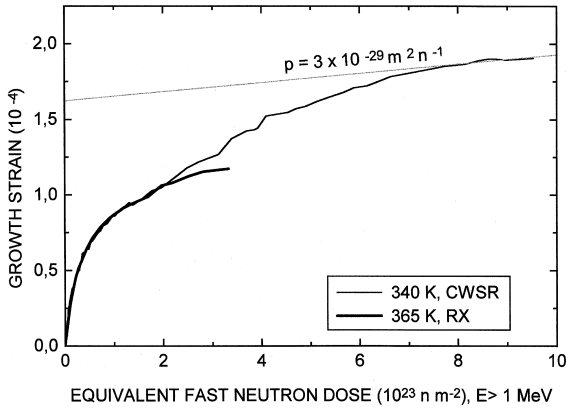


Fig. 2. Irradiation growth of specimens irradiated at CNEA-RA1. In the abscissa, the equivalent dose calculated with the method described in the text. The asymptotic behaviour shows the initiation of the constant deformation stage. $p$ represents the slope of the limiting straight line.

The deformation kinetics of a CWSR material of the same composition irradiated during 300 h at liquid nitrogen temperature is shown in Fig. 3. After a short initial transient, growth saturation is reached at a value of $3.7 \times 10^{-4}$. Once the irradiation was over, the sample was annealed at room temperature during several days and then again measured at 77 K. The deformation recovery was of 85%. Another similar irradiation period showed similar behaviour.

The linearly increasing dependence of growth with dose, characteristic of cold worked and stress-relieved materials, is not observed in the later case. However the saturation dose gave values coincident with those obtained at the other temperatures.

This shows that the initial growth transient is due to the establishment of a distribution of free point defects and clusters of defects that reaches a stationary value in size and distribution after receiving a dose of $1 - 2 \times 10^{23}$ n m$^{-2}$. These clusters are preferred sinks for point defects [31]; then, the growth during this first stage at temperatures close to 350 K is the product of some inferred change in volume, and fundamentally, the preferred collapse of clusters in crystalline plains as dislocations loops with size growing with the dose. After that, the point defects (especially interstitials, which are mobile at the irradiation temperature), continue migrating to the other sinks contributing to the deformation. In the case of the CWSR material, the absorption of defects by loops competes with the previously existent dislocations since they reach a size such that their bias to

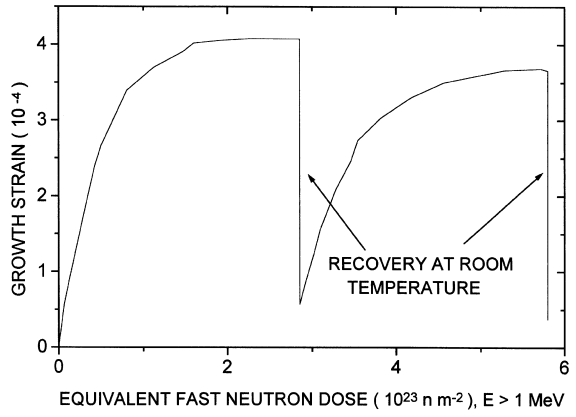


Fig. 3. Measurement of the longitudinal deformation at 77 K.

point defects is similar. Conversely, in RX materials the contribution of dislocations is not significant and consequently the growth saturates or continues at lower rate. For these reasons, the processes in this zirconium alloy with different thermal treatments begin to diverge at doses higher than $1–2 \times 10^{23}$ n m$^{-2}$. This demonstrates that the dislocation loops, which grow by absorption of point defects, constitute the principal contribution to deformation during the initial transient at temperatures around 350 K.

At 77 K the mechanisms are obviously different to those responsible for the dimensional variation at 340 and 365 K. The recovery at room temperature of most of the deformation in a CWSR material is significant. Harbottle [18] also obtained a partial recovery at room temperature. The results of the measured kinetics suggest that a volume change takes place producing most of the deformation, the balance corresponding to the distribution of agglomerates produced in the collision cascade; their size, character and number is different to that obtained at 350 K.

At low temperatures (77 K), many of the point defects contribute to changes of volume; at higher temperatures (Stage III of recovery with mobility of interstitials) the point defects contribute to the growth of dislocation loops, that is, to the irradiation growth. Then, at $1–2\times 10^{23}$ n m$^{-2}$ the defects reach a dinamical equilibrium, but with different distribution at each temperature, and independent of the metallurgical condition of the alloy.

At temperatures >400 K (Stage IV), the mobility of the vacancies increases the i–v recombination. They also flow to others sinks, such as interstitial loops, whose size grow, then, more slowly. For this reason the first stage of the deformation ends at a higher doses [17].

A linear dependence of growth in CWSR zirconium alloys is reported in the literature [32], starting from a dose close to $1–2 \times 10^{24}$ n m$^{-2}$ ($E > 1$ MeV), and also, with a rate ($\dot{\varepsilon}_g$) increasing with the amount of cold work. Fig. 2 shows a slope of $3 \times 10^{-29}$ m$^2$ n$^{-1}$ in the growth curve obtained at 340 K; this is coincident with the growth rate obtained by Murgatroyd and Rogerson [33] at 353 K for Zr alloys at high doses and with 25% of cold work. This indicates that, at equivalent doses $\leqslant 0.7 \times 10^{24}$ n m$^{-2}$, the CWSR material has already initiated the linear growth regime.

## 4. Conclusions

The initial irradiation growth transient was measured at temperatures close to 350 K in two specimens of the same composition, one with recrystallization treatment and the other with 50% cold work and stress-relief. Deformation behavior coincides exactly in both samples up to an equivalent dose of $1 – 2 \times 10^{23}$ n m$^{-2}$. The saturation value of deformation also coincides with that of another specimen irradiated at 77 K. This fact shows the influence of the defects created by irradiation, particularly the loops originated in the displacement cascades.

A dose of $1 – 2 \times 10^{23}$ n m$^{-2}$ is the starting point of the second stage of deformation, where the growth rate depends on the thermal treatment of the material.

It was possible to estimate that at doses as low as $7 \times 10^{23}$ n m$^{-2}$ the long-term linear growth regime is already established in this zirconium alloy.

The coincidence of the slope measured in a CWSR Zr-1.5 at.% Al alloy doped with 0.12 at $^{235}$U with data reported in the literature for other zirconium alloys in similar conditions reveal that the calculation method used to evaluate the equivalent neutron dose is correct within the experimental error of the local neutron flux.

## Acknowledgements

The authors wish to thank Mr C.D. Anello for his technical assistance throughout these experiments. Part of this work was funded by the CANDU Owners Group under WPIR 3202.

## References

[1] R.A. Holt, E.F. Ibrahim, Acta Met. 27 (1979) 1319.
[2] V. Fidleris, R.P. Tucker, R.B. Adamson, in: Zirconium in the Nuclear Industry, ASTM-STP 939, American Soc. for Testing and Materials, Philadelphia, PA, 1987, p. 49.
[3] V. Fidleris, J. Nucl. Mater. 159 (1988) 75.
[4] R.G. Fleck, R.A. Holt, V. Perovic, J. Tadros, in: Proceedings of the International Conference on Fundamental Mechanisms of Radiation-Induced Creep and Growth, Hecla Island, Manitoba, Canada, 22–25 June 1987, J. Nucl. Mater. 159 (1988) 75.
[5] D.O. Northwood, R.W. Gilbert, L.E. Bahen, P.M. Kelly, R.G. Blake, A. Jostons, P.K. Madden, D. Faulkner, W. Bell, R.B. Adamson, J. Nucl. Mater. 79 (1979) 379.
[6] M. Griffiths, J. Nucl. Mater. 159 (1988) 190.
[7] R.B. Adamson, R.P. Tucker, V. Fidleris, in: International Conference on Zirconium in the Nuclear Industry, 1980, Boston, USA, ASTM STP 754, ASTM, Philadelphia, 1982, p. 208.
[8] M. Griffiths, R.W. Gilbert, V. Fidleris, R.P. Tucker, R.B. Adamson, J. Nucl. Mater. 150 (1987) 159.
[9] A. Rogerson, in: Proceedings of the VI International Symposium on Zirconium in the Nuclear Industry, Vancouver, B.C., 28 June–1 July 1982, ASTM STP 824, ASTM, Philadelphia, 1984, p. 394.
[10] A. Rogerson, J. Nucl. Mater. 159 (1988) 43.
[11] A. Rogerson, R.A. Murgatroyd, J. Nucl. Mater. 113 (1983) 256.
[12] R.A. Holt, R.W. Gilbert, J. Nucl. Mater. 137 (1986) 185.
[13] R.A. Holt, J. Nucl. Mater. 159 (1988) 310.

[14] M. Griffiths, R.A. Holt, A. Rogerson, J. Nucl. Mater. 225 (1995) 245.
[15] P.R Tucker, V. Fidleris, R.B. Adamson, in: Proceedings of the VI International Conference on Zirconium in the Nuclear Industry, Vancouver, Canada, 28 June–1 July 1982, ASTM STP 824, ASTM, Philadelphia, 1984, p. 427.
[16] C.H. Woo, Rad. Eff. Def. Sol. 144 (1998) 145.
[17] G.D.H. Coccoz, A.M. Fortis, H.C. González, J. Nucl. Mater. 229 (1996) 73.
[18] J.E. Harbottle, ASTM-STP 484, Irrad. Eff. of a Structural Alloy for Nuclear Reactor Application, Philadelphia, PA, (1970) 287.
[19] J.P. Parsons, C.W. Hoelke, in: Proceedings of the 14th International Symposium Effects of Radiation on Materials. Vol II, Andover, MA, USA 27–30 June 1988. ASTM-STP. 1046, ASTM, Philadelphia, 1990, p. 588.
[20] A.M. Fortis, E. Mezzabolta, H.C. González, Nucl. Instrum. Meth. A 293(3) (1985) 592.
[21] A.M. Fortis, H.C. Gonzalez, Rev. Asoc. Quím. 84 (5) (1996) 523.
[22] G.H. Kinchin, R.S. Pease, Prog. Phys. 18 (1955) 1–51.
[23] W.S. Snyder, J. Neufeld, Phys. Rev. 97 (1955) 1636.
[24] M.J. Norgett, M.T. Robinson, I.M. Torrens, Nucl. Eng. Des. 33 (1975) 50.
[25] J. Lindhard, M. Scharf, H.E. Schiøtt, K. Dan. Vidensk Selsk. Mater. Fys. Medd. 33 (10) (1963) 1.
[26] M.T. Robinson, in: N.L. Petersen, S.D. Harkness (Eds.), Radiation Damage in Metals, ASM seminar, 1975, p. 12.
[27] R.S. Averback, H. Horngming, T. Díaz de la Rubia, R. Benedek, J. Nucl. Mater. 179–181 (1991) 87.
[28] M.J. Norgett, M.T. Robinson, I.M. Torrens, Nucl. Eng. Des. 33 (1975) 50.
[29] M.T. Robinson, J. Nucl. Mater. 216 (1994) 1.
[30] Zou. Yiming, Ma. Gonggui, Wang. Shiming, Comm. Nucl. Data Progr. 6 (1991) 158.
[31] M. Halbwachs, J. Hillairet, H.C. González, J. Cost. Radiat. Eff. 30 (1976) 171.
[32] R.B. Adamson, Zirconium in the Nuclear Industry, ASTM-STP 633, American Soc. for Testing and Materials, Philadelphia, PA, 1977, p. 326.
[33] R.A. Murgatroyd, A. Rogerson, Zirconium in the Nuclear Industry, ASTM-STP 681, American Soc. for Testing and Materials, Philadelphia, PA, 1979, p. 213.