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Mechanical and Constitutive Behaviour of Zr-2.5Nb Pressure Tube Material at High Rates of Tensile Strain

by

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B.Sc. (Applied Mathematics and Physics)

M.Sc. (Mechanical Engineering)

A thesis submitted to
the Faculty of Graduate Studies and Research
in partial fulfilment of the requirements
for the degree of

Doctor of Philosophy
in Mechanical Engineering

Ottawa-Carleton Institute
for Mechanical and Aerospace Engineering
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and Research acceptance of the thesis

**Mechanical and Constitutive Behaviour of Zr-2.5Nb Pressure
Tube Material at High Rates of Tensile Strain**

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In partial fulfillment of the requirements for the degree of Doctor of Philosophy

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Abstract

A direct-tension split Hopkinson bar apparatus was used to perform experiments on Zr-2.5Nb pressure tube material at tensile strain rates in the range of 500 to 3000 s⁻¹, temperatures between 20 and 300°C, and fluences up to 8 x 10²⁵ n/m². Tests were performed on specimens manufactured from both the longitudinal and transverse tube orientations. Experimental results on non-irradiated material showed that the effect of increasing the rate of strain was to increase the strength and decrease the ductility for both specimen orientations. The effect of irradiation was to further increase the strength and decrease the ductility. All of the irradiated material, however, exhibited a significant amount of plastic deformation and failed in a ductile manner. A bi-linear dependence of flow stress on log-strain rate exists for both the non-irradiated and irradiated material, with the rate sensitivity being greater over the high strain rate regime. Zr-2.5Nb pressure tube material remains anisotropic at high strain rates with the strength being greater in the transverse orientation. The effect of increasing the temperature was to decrease the strength over the range of strain rates and fluences considered.

Results from the mechanical tests were used to characterise the constitutive behaviour in the longitudinal direction with the Johnson-Cook and Zerilli-Armstrong hcp relations. Both constitutive relations closely predict the behaviour of the non-irradiated material for the range of strain rates and temperatures investigated, if the effects of localised deformation and adiabatic heating were incorporated in the derivation of the constitutive parameters. The irradiated responses, however, were not as closely predicted

due to the relatively large variability in the data for the irradiated samples. The Zerilli-Armstrong hcp relation produced slightly more accurate results for the conditions considered.

Finite element simulations of the direct-tension split Hopkinson bar experiments were performed using the derived constitutive relations to describe the material behaviour. Results from the simulations showed that the finite element method can be used to predict the longitudinal stress-strain behaviour of as-received and irradiated Zr-2.5Nb pressure tube material at high rates of tensile strain. The onset of a diffuse necked region and the reduction in cross-sectional area to the point of failure were also predicted with reasonable accuracy. The anisotropic deformation behaviour exhibited by the radial and transverse strains in the necked region, however, was not captured.

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Chapter 1

Introduction

It is well established that most metals and alloys show a significant change in mechanical behaviour under elevated strain rates [1 - 8] and for approximately 50 years experimental techniques have been advanced to characterise materials under dynamic or high strain rate loading conditions [4, 5, 8, 9]. As a result, the behaviour of a wide variety of engineering materials have been characterised, allowing for the development of constitutive relations capable of accurately predicting the deformation of materials. The availability of modern high-speed computers along with these sophisticated constitutive relations makes it possible to analyse complex problems such as automotive crash worthiness, foreign object damage, metal forming, ballistic penetration, and postulated accident scenarios in nuclear reactors.

The objective of the current research is to characterise the high strain rate tensile properties of non-irradiated (as-received) and irradiated Zr-2.5wt.%Nb (hereafter denoted as Zr-2.5Nb) pressure tubes used in Canada Deuterium Uranium (CANDU[®]) nuclear

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reactors and to develop constitutive relations that can be used in finite element codes for the analysis of postulated accident scenarios involving pressure tube rupture.

The following sections of this chapter review the main topics related to this thesis, starting with Section 1.1, which provides an overview of the problem being addressed. A detailed description of the material used for this research is given in Section 1.2. Next, the behaviour of metals and alloys under high rates of strain (Section 1.3), and the current constitutive relations used to describe deformation behaviour during high strain rate loading (Section 1.4) are described. Finally, a review of the experimental techniques used to characterise materials under high strain rate loading conditions is provided in Section 1.5.

1.1 Motivation

The current design of CANDU reactors utilise a large number (~ 400) of individual fuel channels that are surrounded by a heavy water (D_2O) moderator and are contained within a large vessel known as the calandria, Figure 1.1. As shown in Figure 1.2, each fuel channel consists of an outer tube (calandria tube) and an inner pressure flow tube (pressure tube) that contains the uranium fuel bundles and the heavy water coolant. The pressure tubes are manufactured from cold-worked Zr-2.5Nb material [10] and are required to operate at internal pressures up to 13 MPa, temperatures in the range of 250 to 310°C, and a fast neutron flux of 2.5×10^{17} n/m²/s, with neutrons having an energy in excess of 1 MeV ($E > 1$ MeV). The CO₂ gas annulus and surrounding

calandria tube act to thermally insulate the pressure tube from the moderator, which under operating conditions is at a temperature of approximately 80°C.

In certain hypothetical accident scenarios, during the normal operation of a CANDU reactor, a pressure tube could burst prior to the detection of a through-wall axial crack in the tube and shut-down of the reactor. Assuming also, as a worst case scenario, that the surrounding calandria tube fails, the longitudinal crack in the pressure tube could continue to propagate down the length of the tube. In this situation the pressure tube would deflect vertically due to the release of the pressurized coolant, thereby dynamically loading the ends of the pressure tube in tension. Depending on the final crack geometry and thermal-hydraulic conditions, localised deformation at the ends of the pressure tube could cause the formation of a complete circumferential crack, or guillotine failure [11] resulting in further in-core damage and end-fitting ejection. Preliminary analysis [12] indicated that plastic deformation of a rupturing pressure tube in the longitudinal direction, with no support of a calandria tube, would occur at strain rates up to approximately 3000 s^{-1} . Prior to the current research program, there was no high strain rate data available for Zr-2.5Nb pressure tube material. Finite element analysis of hypothetical accident scenarios involving pressure tube rupture relied on existing quasi-static data and an assumed elastic, perfectly-plastic work hardening relationship to describe material deformation [12]. Because of the documented effects of high strain rate loading conditions on the mechanical response of other metals and alloys [1 - 8], constitutive relations describing the flow stress as a function of strain, rate of strain, and temperature for Zr-2.5Nb pressure tube material are required. Since the end objective was to apply the constitutive relations in finite element calculations of

postulated pressure tube rupture scenarios, these constitutive relations would have to be compatible with the typical continuum mechanics approach of the finite element method.

1.2 Zr-2.5Nb Pressure Tube Material

Zirconium is used as the base alloy in pressure tubes because it has good mechanical and corrosion resistant properties, can be manufactured into tubes, and has a low thermal neutron capture cross section [13]. The design is based on Section III of the American Society of Mechanical Engineers (ASME) Pressure Vessel Code. The restrictions on the maximum design stress at operating temperatures are one-third of the ultimate tensile strength (σ_{UTS}), or two-thirds of the 0.2% offset yield strength (σ_y). For Zr-2.5Nb pressure tubes, one-third of the σ_{UTS} is the limiting property [10]. The fabrication process of pressure tubes is rigidly controlled and they are required to meet stringent chemical and geometric specifications [10]. This section provides an overview of the fabrication process of Zr-2.5Nb pressure tubes, the resulting microstructure and deformation mechanisms, effects of irradiation, and mechanical properties relevant to the current research.

1.2.1 Fabrication

The overall fabrication process of Zr-2.5Nb pressure tubes is summarised in Figure 1.3 [10] and can be divided into five main steps [14]: production of ingots,

production of billets, extrusion of billets into tubes, cold reduction, and heat treatment. The Zr-2.5Nb alloy is made by arc melting zirconium sponge and a master alloy of high-purity zirconium and niobium. This process produces an ingot nearly free of inclusions with a maximum total impurity content of approximately 3400 ppm [10]. The ingots are then preheated to approximately 980°C and forged into round bars, using flat and rotary forges [14], which are machined to produce hollow billets. In order to produce a more uniform microstructure, the billets are beta quenched from 1020°C [10], and then clad in copper sheet to minimise oxidation [15]. The extrusion process is performed at approximately 850°C using a press forge at an extrusion ratio of 10.5:1 to produce an extruded tube approximately 5.8-m-long with a wall thickness of 6 mm and an inside diameter of 107 mm. The as-extruded tubes are then cold drawn approximately 26% to the final tube dimensions of 6.3 m in length, 4.2 mm wall thickness ($4.191 + 0.483 - 0.000$ mm), and an inside diameter of 103 mm ($103.378 + 0.838 - 0.000$ mm), using a plug drawing process [10]. After the cold reduction process, the pressure tubes are given a final heat treatment which is performed in a steam autoclave at 400°C for 24 hours. This treatment serves to stress relieve the tube and form a hard, protective oxide layer [10].

1.2.2 Microstructure and Crystallographic Texture

Zirconium undergoes the allotropic transformation [16] from hexagonal-close-packed (hcp) zirconium (α -zirconium) to body-centred-cubic (bcc) zirconium (β -zirconium) upon heating above a temperature of 862°C, as illustrated in Figure 1.4. The principal crystallographic planes for α -zirconium include the basal

planes (0001), prism planes (01 $\bar{1}$ 0), and two types of pyramidal planes (1 $\bar{1}$ 01), (11 $\bar{2}$ 1), as shown schematically in Figure 1.5.

Because of the fabrication process of Zr-2.5Nb pressure tubes, a very distinct grain structure and texture is developed from the two-phase alloy [10]. At equilibrium below 600 °C, the α -zirconium phase (α -phase) can contain approximately 0.6 wt.% niobium in solid solution. The remaining 1.9 wt.% niobium is present within the cubic β -zirconium phase (β -phase). At the extrusion temperature of 850°C, the microstructure consists of approximately 20% α -phase and 80% β -phase. The effect of extrusion is to elongate the alpha grains so that they are long and thin. During air cooling, the β -phase transforms to α -phase by growth on the existing alpha grains. This results in the β -phase becoming progressively enriched in niobium until it contains 18 to 20% niobium. The β -phase with this composition is metastable at room temperature. Therefore, the microstructure of the pressure tubes at the “as-extruded” stage consists of alpha grains up to 20- μ m-long, 2- μ m-wide and 0.4- μ m-thick, with a surrounding grain boundary network of β -phase that contains 18 to 20% niobium [10]. Upon cold reduction of the pressure tube wall thickness by 25 to 30%, the α -phase grains become more elongated in the longitudinal direction of the tube. Figure 1.6a shows a micrograph of the final as-received pressure tube material (x 1000 mag.) taken in the longitudinal-radial plane. The elongated α -phase grains are aligned in the longitudinal direction. The final aspect ratio of the α -phase grains is approximately 1:5:50 for the radial, transverse and longitudinal directions, respectively. The dimension of the α -phase grains in the radial direction typically varies from approximately 0.3 to 0.5 μ m for as-received material, with

the occasional grain as large as $2\text{ }\mu\text{m}$ [10]. Figure 1.6b shows a micrograph of the transverse-radial plane of as-received pressure tube material.

Alpha-zirconium has a hexagonal crystal structure and a limited number of deformation systems [16-26]. This causes the α -phase grains of Zr-2.5Nb pressure tube material to form a characteristic crystallographic texture during fabrication [10, 24, 27]. The final texture of the pressure tubes is such that most of the grains are oriented with their $(10\bar{1}0)$, $(11\bar{2}0)$, and (0001) planes perpendicular to the longitudinal, radial, and transverse directions, respectively. Pole figures for the (0001) and $(1\bar{1}00)$ planes are given in Figure 1.7 and an idealised representation of the texture in the tube is given in Figure 1.8.

1.2.3 Deformation Mechanisms

The deformation mechanisms of polycrystalline Zr-2.5Nb pressure tube material are closely represented by those of single crystal α -zirconium [10, 13, 24] due to the extent of texture developed during fabrication. Generally, the more pronounced the texture of a polycrystal, the closer the properties of the single crystal are emulated [26]. The deformation systems in low symmetry materials, such as hcp α -zirconium, are not as symmetrically distributed as in cubic materials and are generally more complicated [16, 25, 26]. Overall, there are fewer primary slip systems available in hcp materials and they tend to be associated with deformation in the a -direction. Deformation resulting from twinning in materials often competes with that of slip and is used to explain

deformations with c -components. As a result, twinning plays a major role in the plastic deformation of zirconium and Zr-2.5Nb pressure tube material [13].

The slip systems in zirconium have been examined over a temperature range of -200 to 800°C by a number of investigators [17 - 23, 25, 28 - 39] and reviewed in [16, 26]. The general consensus from these investigations is that the slip systems in α -zirconium are limited, with the major system in the temperature range of room temperature to approximately 500°C being slip on $\{10\bar{1}0\}$ first-order prism planes in the $\langle 1\bar{2}10 \rangle$ a -direction [16, 18, 19, 21, 26, 28, 30 - 35], as shown schematically in Figure 1.9 a. Slip was also observed [16, 21, 26, 35, 36] on the (0001) basal plane in the same direction, as shown in Figure 1.9 b. In regions of high stress concentration, $\{10\bar{1}1\}$ slip was reported to occur [16, 18, 26, 34, 37]. Slip with a c -component was only observed under conditions of high constraint and at elevated temperatures [16, 26, 29, 38, 39]. This occurred on first and second-order pyramidal planes, $\{10\bar{1}1\}$ and $\{11\bar{2}1\}$, in a $(c + a)$ direction, as shown in Figure 1.9 c. Further slip planes were reported, but they did not have well defined directions. These included $\{10\bar{1}2\}$ and $\{10\bar{1}3\}$ slip reported by [32], and $\{11\bar{2}2\}$ slip reported by [18, 34].

In zirconium, the overall magnitude of deformation due to twinning is generally small relative to that caused by slip. However, twinning is essential to ductility because the orientation change in the lattice may be large, thereby re-establishing orientations that are favourable for slip. Because of its importance in the deformation of zirconium, twinning has been fairly well studied and documented [16 - 23, 25, 26, 30, 31, 37, 40 - 48]. The general consensus in the literature is that the main twin planes in α -zirconium

are $\{10\bar{1}2\}$, $\{11\bar{2}1\}$ and $\{11\bar{2}2\}$. Also, twinning along the $\{11\bar{2}3\}$ planes occurs, but is usually not included as a major twinning system because it requires very high strains [16, 26]. The twin systems activated for tensile loading along the c -axis, or $\langle 0001 \rangle$ direction, are $\{10\bar{1}2\} \langle \bar{1}011 \rangle$ [16, 18, 19, 21, 26, 30, 31, 37, 41 - 43, 46, 47], and less commonly $\{11\bar{2}1\} \langle \bar{1}\bar{1}26 \rangle$ [16, 18, 21, 26, 30, 31, 33, 37, 42 - 47], as shown in Figure 1.10a. Under compression along the c -axis, $\{11\bar{2}2\} \langle \bar{1}\bar{1}23 \rangle$ twinning [16, 18, 19, 21, 26, 31, 33, 34, 37, 42, 43, 46, 47], and at elevated temperatures $\{10\bar{1}1\} \langle \bar{1}012 \rangle$ twinning [16, 26, 38, 43, 46, 47] were reported, as shown in Figure 1.10b. Limited twinning on the $\{11\bar{2}3\}$ compound plane was reported [16, 18, 19, 21, 26, 30, 31], but they were not completely defined.

In order to predict which deformation mode will be activated, consideration must be given to the geometrical relationship between the direction of applied force and the corresponding deformation mode [26], as shown schematically in Figure 1.11 and represented by,

$$\tau_{RSS} = \frac{F}{A} \mu \quad (1.1)$$

where τ_{RSS} is the resolved shear stress, F is the externally applied force, A is the cross-sectional area of the crystal, and μ is the Schmid Factor [26]. The Schmid Factor is further defined as,

$$\mu = \cos\phi \cos\lambda \quad (1.2)$$

where for slip, τ_{RSS} is the resolved shear stress on the slip plane in the slip direction, ϕ is the angle between the normal to the slip plane and the direction of the force, and λ is the angle between the slip direction and the direction of the force. For twinning, τ_{RSS} is the resolved shear stress on the twinning plane in the shear direction, ϕ is the angle between the normal to the twinning plane and the direction of the force, and λ is the angle between the shear direction of twinning and the direction of the force. For a given crystallographic orientation, a material will initiate deformation by slip if τ_{RSS} is greater than a critical activation stress associated with the particular slip system, known as the critical resolved shear stress, τ_{CRSS} . Conversely, if the stress required to initiate deformation twinning is lower than τ_{CRSS} for the available slip systems, the material will initiate deformation by twinning. The critical resolved shear stress associated with slip is generally lower than the stress required to initiate twinning, but is affected by such factors as temperature, strain rate, and chemistry [26, 47].

The deformation system with the lowest activation stress, however, will not necessarily be operative. The availability of slip and twinning systems for a given crystallographic orientation, relative to the direction of stress, must be considered. For zirconium, the resolved shear stress associated with the various slip and twin systems are strongly dependent on the direction of force, especially with respect to the *c*-axis, or basal pole [21, 26]. This dependence is shown schematically for some of the major slip and twinning modes in Figure 1.12 for a range of angles from 0 to 90° between the basal pole

and the direction of force. In Figure 1.12 the region labelled prism slip corresponds to the prism plane, $\{10\bar{1}0\}$, that is at the highest level of stress for a given angle. The region labelled $\{10\bar{1}2\}$ twin, represents the range of the orientation factors for all six $\{10\bar{1}2\}$ twinning planes. The interpretation of Figure 1.12 [21] is that when the direction of stress, or axis of loading, approaches the basal pole, the resolved shear stress on the prism slip plane is low and the corresponding stress associated with the six $\{10\bar{1}2\}$ twinning planes becomes large. Therefore, prism slip and $\{10\bar{1}2\}$ twinning can be viewed as being interdependent deformation systems in zirconium [21, 26].

As described in Section 1.2.2 and shown in Figure 1.8, Zr-2.5Nb pressure tube material is highly textured resulting in the majority of the α -zirconium grains having the c -axis oriented in the transverse tube direction. The deformation system that is activated depends primarily on the crystallographic orientation of the c -axis with reference to the direction of applied load, or the Schmid factor [26]. For Zircaloy[®] tubing [49], with a similar texture as Zr-2.5Nb pressure tube material, it was shown for tensile loading in the axial tube direction that the primary deformation mode was $\{10\bar{1}0\} \langle 11\bar{2}0 \rangle$ prismatic slip in the longitudinal direction and $\{11\bar{2}2\}$ twinning in the transverse direction. These results compared well with predicted behaviour and the tube was observed to accommodate deformation through reduction in the wall thickness (radial direction) with little change in diameter (transverse direction). Results from a number of other experimental investigations on textured Zircaloy-2[®] sheet [50], Zircaloy-2[®] and Zircaloy-4[®] tubing [27, 51 - 53], and Zr-2.5Nb tube material [13, 27, 54] agree with the

[®] Zircaloy is a trademark of Westinghouse Electric Company, Pittsburgh, PA.

theoretically derived behaviour predicted in [26]. Following these observations, it is expected that the Zr-2.5Nb pressure tube material used in the current investigation will deform primarily by $\{10\bar{1}0\} \langle 11\bar{2}0 \rangle$ prismatic slip when loaded in tension in the longitudinal tube direction and $\{11\bar{2}2\}$ twinning when loaded in tension in the transverse tube direction.

Further to the major effect of crystallographic orientation on the operative deformation mechanisms, the effects of temperature, strain rate and chemistry are of interest in the present investigation since Zr-2.5Nb pressure tube material contains over 3400 ppm of impurities and tests at high strain rates are conducted. The effects of strain rate and temperature on materials are generally related through an Arrhenius type equation [47]. The mechanisms controlling slip are sensitive to changes in temperature and therefore, changes in the rate of strain. As the temperature is decreased, or strain rate increased, the τ_{CRSS} for slip systems increases. Since the magnitude of the stress required to initiate twinning is less temperature dependent, there is a tendency to substitute deformation due to slip with twinning with decreasing temperature [21, 55]. Recent tests on the hcp titanium alloy, Ti-6Al-4V [56, 57], pure titanium [58], and zirconium [59, 60] have further confirmed this tendency by demonstrating that the propensity for twinning increased as the deformation temperature decreased, or the rate of strain was increased [59 - 61]. The effect of relatively small levels of impurities (~300 ppm) in pure zirconium was found to suppress the formation of twins, even at low temperatures [60, 61]. This is consistent with a number of other results on bcc metals and alloys [47].

1.2.4 Irradiation Damage

The effect of irradiation of materials has received considerable attention since 1946 [13], when Wigner [62] demonstrated that an energetic particle striking an atom in a crystalline matrix could impart sufficient energy to displace the atom from its original position. Because of the extensive use of zirconium for fuel cladding and pressure tubes in the nuclear industry, a significant amount of work has involved zirconium alloys. The majority of the literature on zirconium alloys, however, relates the effects of irradiation damage to mechanical properties. This section provides a short overview of the general concepts of irradiation damage on materials. The effects on the mechanical properties of Zr-2.5Nb pressure tube material are discussed in Section 1.2.5.

Following the theoretical analysis presented in [63], the first atom in a lattice to receive the energy from an impinging particle is called the primary knock-on atom (PKA). If the PKA energy is greater than the threshold displacement energy, then the atom is displaced from its lattice position leaving a vacancy and forming an interstitial atom in the lattice where it comes to rest. This combination of the vacancy and the PKA is known as a Frenkel Pair. If the PKA has sufficient energy, it may collide with and displace other atoms, forming additional Frenkel Pairs. This process, known as a collision cascade, will continue until the energy has been dissipated. Calculations presented in [63] showed that 600 to 1000 Frenkel Pairs are created in a collision cascade caused by a single impinging neutron with energy greater than 1 MeV. Alternatively, if the PKA energy is less than the threshold displacement energy, the PKA will recoil back into its original position within the lattice.

The geometrical distribution of atoms after radiation induced damage was first addressed by Brinkman [64]. The basic assumption was that essentially every atom in the path of the impinging particle is displaced. In turn, a number of these atoms may displace others, and so on. The density of the damage then increases as the energetic particle loses energy, producing a region within the lattice where all of the atoms are displaced. The distance that the atoms are displaced is short, but together they form what Brinkman [64] called a *Displacement Spike*, as shown schematically in Figure 1.13. The interstitials form a shell around a large vacancy, or hole. Because of the displacement of the atoms from their original positions in the lattice, there is a net increase in temperature, above the melting temperature, denoted as a *Thermal Spike*. The interstitials can then diffuse back to the original lattice positions, resulting in little permanent damage to the lattice.

This model was refined by Seeger [65], and is schematically represented in Figure 1.14. In this model, the assumption is that upon relaxation of the region, interstitials far enough away from the spike do not feel the strain field, and therefore, do not return to the original positions. This results in a region at the center of the collision cascade that is missing atoms. This region is called a *depleted*, or *Seeger zone*, and is typically < 1 nm in size. Under the appropriate conditions, defects due to radiation can form interstitial or vacancy dislocation loops [13], as shown in Figure 1.15.

The exposure of zirconium alloys to fast neutron irradiation produces defects in the α -zirconium phase of size and distribution that depend on neutron fluence, irradiation temperature, and alloying content [13, 16, 66, 67]. For neutron fluences of approximately $2 - 3 \times 10^{23} \text{ n/m}^2$, it was reported [13, 66, 67] that there was no noticeable

damage to zirconium alloys, in the form of detectable dislocation loops. At fluences $> 3 \times 10^{23} \text{ n/m}^2$, dislocation loops occurred and their density increased for increasing fluence, as did the defect size. It was observed, however, that the number of atom sites involved in the damage saturated at a fluence of approximately $1 \times 10^{25} \text{ n/m}^2$.

The overall effect of increasing the irradiation temperature is to cause a decrease in the number of defects, but increase their size [13, 16]. As the irradiation temperature is increased above approximately 300°C, there is less irradiation induced hardening. Also, there is a threshold irradiation temperature for damage, which depends on alloy content and flux. Above the threshold temperature, irradiation induced damage is annealed out at the same rate that it is created and essentially no net damage results.

A number of investigations into the characteristics of irradiation induced damage in zirconium alloys [13, 67 - 71] show that $a/3 \langle 11\bar{2}0 \rangle$ is the most likely Burgers vector. There was general agreement that these $\langle a \rangle$ -type dislocations lie in or near the prism planes, and were generally a mixture of interstitial and vacancy loops. Damage associated with $\langle c \rangle$ -type dislocations was less prevalent [69]. This results in deformation by prismatic slip, which is associated with $\langle a \rangle$ -type defects being affected to a greater extent than other deformation mechanisms involving $\langle c \rangle$, or $\langle a \rangle$ and $\langle c \rangle$ -type defects [72].

A recent study on the effect of long-term irradiation on the properties of Zr-2.5Nb pressure tube material demonstrated that neutron irradiation under reactor typical conditions ($E > 1 \text{ MeV}$) results in the formation of a high density of $\langle a \rangle$ -type dislocations loops [73]. The density of these loops increased rapidly up to low fluences $< 1 \times 10^{25} \text{ n/m}^2$. Subsequent irradiation lead to little further change in the $\langle a \rangle$ -type

dislocation structure. However, $\langle c \rangle$ -type dislocation structures were observed to form at higher fluence levels and continued to evolve up to fluences of approximately $17 \times 10^{25} \text{ n/m}^2$ [73].

1.2.5 Tensile Properties

The majority of the available information on the deformation of zirconium materials is restricted to Zircalloys®, plate material with very different textures, and specimens that were irradiated to very low fluence levels compared to that expected over the service-life of a pressure tube [74]. Overall, the number of investigations specifically related to the tensile properties of Zr-2.5Nb pressure tube material is quite limited, particularly in the longitudinal direction [75]. Data on pressure tube material has generally been obtained for studies related to characterising the longitudinal fracture toughness [73, 76], and therefore, involves the transverse tensile properties.

As discussed in Section 1.2.4, radiation induced damage in Zr-2.5Nb pressure tube material results in the formation of a high density of defects. Upon deformation of the irradiated material, these defects form obstacles to the glide motion of dislocations. The net effect is to increase the stress at which the onset of free dislocation motion occurs (yield stress) [77]. The increase in the yield stress due to irradiation in a number of metals and alloys is seen to increase proportionally with $(\text{dose})^P$ where, dose represents the total amount of irradiation, measured in neutrons per unit area, and P ranges from 1/3 to 1/2 [77]. Although the above power-law function fits data for a number of metals [77], it was demonstrated by Bement [67] that such a function was only applicable for

relatively low doses, or total fluences of approximately $6 \times 10^{22} \text{ n/m}^2$. It was further demonstrated [67] that an inverse exponential function, such as the one introduced in [78], better represented yield stress data obtained at higher fluence levels, where damage due to radiation is known to asymptotically reach a maximum value.

As the obstacles opposing dislocation motion are overcome, the local flow stress is immediately lowered causing an avalanche of instability that clears out an obstacle-free path, or channel [64, 77, 79]. This phenomenon has been used to explain the occurrence of a yield drop in a number of irradiated materials. Subsequent deformation takes place in these channels that are essential in a “pre-irradiated” state, and therefore, the deformation behaviour is similar to that observed in non-irradiated material [64, 77, 79]. As a result of this strain localisation, there is generally a significant reduction in both the uniform elongation and strain to fracture in irradiated materials. It should be noted, however, that irradiated Zr-2.5Nb pressure tube material fails in a ductile manner [80].

Prior to the testing of actual Zr-2.5Nb pressure tubes irradiated in a reactor [81], two investigations were conducted on experimental pressure tubes [82, 83]. These tubes were manufactured from cold-worked Zr-2.5Nb material and had similar geometries, microstructures and chemical compositions to that of actual pressure tubes. The experimental tubes had been irradiated in research reactors to relatively low fluence levels in the range of 6.5×10^{23} to $8.5 \times 10^{24} \text{ n/m}^2$ ($E > 1 \text{ MeV}$). Longitudinal and transverse tensile specimen were manufactured and tested at temperatures ranging from 20 to approximately 300°C and strain rates of approximately $3 \times 10^{-4} \text{ s}^{-1}$. Because of the curvature of the pressure tube in the transverse direction the transverse tensile tests were performed on circular, cross-sectional rings of pressure tube or on typical dog-bone style

specimens that were manufactured from flattened pressure tube. Accurate measurements of the strain were not possible, however, since a gauge length could not be defined for the ring specimens and flattening caused an unknown amount of plastic strain. The longitudinal specimens were manufactured as typical dog-bone tensile specimens with a 25.4 mm gauge length and the strain was measured using output from the cross-head displacement.

The results from these investigations [82, 83] indicated that the pressure tube material was stronger in the transverse direction, consistent with the behaviour expected based on the crystallographic textures. That is, it was expected that due to the distinct crystallographic structure of the material, deformation would take place primarily by slip in the longitudinal direction and twinning in the transverse. This resulted in the longitudinal and transverse specimens exhibiting quite different stress-strain responses [82]. The longitudinal specimens had a lower yield point than the transverse specimens and showed a significant amount of work hardening before the σ_{UTS} was reached. Conversely, the transverse specimens showed very little work hardening with the σ_{UTS} occurring at low plastic strains, close to the σ_y .

Irradiation increased the σ_y and σ_{UTS} , and decreased the ductility over the range of temperatures tested, for both specimen orientations [82, 83]. However, there were larger increases in σ_y and σ_{UTS} for the longitudinal specimen, which was attributed to irradiation damage having a greater effect on the slip systems as compared to the twin systems [82].

In 1987, Ibrahim [81] investigated the effects of irradiation on the tensile properties of actual Zr-2.5Nb pressure tubes by performing tests on specimens manufactured from as-received material and pressure tubes that operated in a reactor for

approximately 12 years (max. fluence $\sim 8 \times 10^{25} \text{ n/m}^2$) at temperatures of approximately 270°C. The tensile tests were conducted at a strain rate of $8.3 \times 10^{-4} \text{ s}^{-1}$ until σ_y was reached and then $8.3 \times 10^{-3} \text{ s}^{-1}$ until failure. As in the previous investigations [82, 83], ring specimens were used for tests in the transverse direction and typical dog-bone style specimens for the longitudinal direction. Also, the strain was calculated from the cross-head displacement.

Results from tests on the as-received material at 300°C indicated that the tubes were an average of approximately 10% stronger in the transverse direction, based on σ_{UTS} values [81]. The σ_{UTS} of the as-received pressure tube material was $549 \pm 30 \text{ MPa}$ for the longitudinal direction and $604 \pm 30 \text{ MPa}$ for the transverse direction. The σ_{UTS} as a function of temperature for one of the pressure tubes tested (fluence $\sim 1.27 \times 10^{25} \text{ n/m}^2$) is shown in Figure 1.16. Included in Figure 1.16 is the effect of temperature on the σ_{UTS} for the as-received pressure tube material. Overall, the results presented by Ibrahim [81] agree with those on the experimental tubes [82, 83] and indicate that irradiation increases the σ_{UTS} in both the longitudinal and transverse directions. Based on σ_{UTS} values, the pressure tube material was strengthened an average of 39% in the longitudinal direction and 23% in the transverse direction at a test temperature of approximately 300°C. The effect of temperature was to decrease the strength for both as-received and irradiated material.

The longitudinal (axial) and transverse σ_{UTS} of Zr-2.5Nb pressure tube material, as a function of fluence, is shown in Figure 1.17. From the figure it is seen that there was a large increase in σ_{UTS} for both directions, to a fluence of approximately $0.7 \times 10^{25} \text{ n/m}^2$, after which there was no further significant increase as the fluence increased to

approximately $8 \times 10^{25} \text{ n/m}^2$. This behaviour indicates that there was a large increase in the σ_{UTS} of the pressure tube material during the first year in-reactor and then little, or no further increase between years 1 and 12 [81].

The total elongation and reduction in cross-sectional area at failure were also measured and the results were used to determine the effect of irradiation on the ductility of Zr-2.5Nb pressure tubes [81]. The results from as-received and irradiated specimens are plotted in Figure 1.18, and show that irradiation reduced the ductility as much as 2 times, at temperatures below approximately 375°C. Above this temperature the reduction in the ductility was marginal. There was little effect of temperature on elongation of as-received and irradiated longitudinal specimens for temperatures below approximately 400 and 300°C, respectively. Above these temperatures, the elongation at failure increased rapidly. It was concluded that irradiation damage was being annealed out over the higher temperature range [81]. The observed temperature dependence of the tensile properties, as shown in Figures 1.16 and 1.18, indicated that the effect of temperature was at a maximum in the 225 to 325°C range. It was concluded [81] that irradiation damage is most effective at impeding deformation at the temperature which it is formed.

The reduction in cross-sectional area at failure, as a function of fluence, is shown in Figure 1.19 for a test temperature of approximately 300°C. It was concluded that there is a large decrease in the ductility as the fluence increased to approximately $0.7 \times 10^{25} \text{ n/m}^2$. For higher fluence levels to approximately $8 \times 10^{25} \text{ n/m}^2$, there was no further decrease to ductility [81].

In more recent work [73], the transverse tensile properties of Zr-2.5Nb pressure tube material were investigated to higher fluence levels ($\sim 17 \times 10^{25} \text{ n/m}^2$). Results showed that there was initially a large increase in strength before a fluence of $0.6 \times 10^{25} \text{ n/m}^2$ followed with a small, but measurable increase to a fluence $17 \times 10^{25} \text{ n/m}^2$. The initial large increase in strength corresponded with an initial rapid increase in $\langle a \rangle$ -type dislocations density as a result of irradiation damage. A slow, but steady increase in $\langle c \rangle$ -type dislocations over the entire fluence range was concluded to be the reason for the observed strengthening over the fluence range of 0.6 to $17 \times 10^{25} \text{ n/m}^2$ [73].

In 1994, the deformation behaviour of irradiated Zr-2.5Nb pressure tube material was investigated by performing tensile tests on specimens manufactured from the longitudinal and transverse directions at temperatures ranging from room temperature to approximately 300°C and a strain rate of $1 \times 10^{-3} \text{ s}^{-1}$ [74]. The material used in this investigation had been in service for up to 15 years in-reactor and had received total fluences in the range of 2.8 to $9.8 \times 10^{25} \text{ n/m}^2$ over a temperature range of approximately 250 to 290°C . Both the longitudinal and transverse specimens were manufactured by first milling a 2-mm-thick flat blank from the thickness of the tube and then spark-machining to the final dimensions. The specimens were approximately 45-mm-long with gauge lengths of 10 and 12.7 mm, respectively, for the transverse and longitudinal specimens. For both specimen orientations, the cross-head displacement was used to determine the strain history.

The overall stress-strain behaviour, as a function of specimen orientation, was consistent with the results observed in previous work [82, 83] and was attributed to the

expected effect of texture on operative deformation mechanisms [74]. That is, the longitudinal specimens are oriented favourably for prismatic slip and the transverse specimens for twinning. As shown in Figure 1.20, the longitudinal specimens were not as strong as the transverse specimen, but exhibited a significant amount of work hardening in comparison. The longitudinal specimens were reported to display uniform elongation throughout the gauge length until a diffuse necked region formed, with eventual failure due to tensile instability, independent of temperature. The transverse specimen, however, showed little work hardening capacity and uniform elongation. As seen in Figure 1.20, the transverse specimens tested above 100°C had a noticeable variation in flow behaviour after the point of σ_y . Below 100°C, all of the transverse specimens exhibited flow behaviour as depicted by curve B in Figure 1.20 [74].

The effect of temperature on σ_y and uniform elongation for longitudinal and transverse specimens manufactured from one tube [74] is shown in Figure 1.21. From the figure it is seen that the σ_y for both specimen orientations decreased for an increase in temperature and that the elongations are nearly independent of temperature [74]. The uniform elongations of the longitudinal and transverse specimens were approximately 3% to 4%, and 0.2% to 0.3%, respectively.

The inconsistent deformation behaviour of the transverse specimen was attributed to the formation of localised deformation bands, in which all of the plastic deformation occurred [74]. The development of deformation bands in irradiated zirconium alloys leading to the localisation of strain have been observed by a number of investigators and were formed as a result of mobilised dislocations annihilating radiation induced defects [67, 74, 79]. By annihilating defects, the stress required to move subsequent dislocations

is reduced, thereby reducing the local flow stress. However, it is believed that twinning rather than dislocation channelling is responsible for the strain localisation observed in irradiated Zr-2.5Nb pressure tube material loaded in the transverse orientation [74]. Since the material is so highly textured, it can behave as a pseudo-monocrystal and twinning can take place across a large number of grains with similar orientation [74].

In 1991, Lam and Shewfelt [75] reviewed the effects of irradiation on the mechanical properties of pressure tube material and reported that there was very little data on the tensile properties of Zr-2.5Nb pressure tube material at strain rates above quasi-static conditions ($\sim 10^{-4} \text{ s}^{-1}$). The only data reported [75] was based on limited tests performed using an instrumented Charpy impact tester [84]. Although the name of the Charpy impact technique implies that it measures the “impact” properties of materials, it should be noted that the technique does not provide a constant strain rate during the test, or throughout the specimen, and the average strain rate is approximately 80 s^{-1} [75]. Data obtained with a Charpy impact machine, therefore, should not be interpreted as typical impact data that would be obtained from high strain rate testing techniques that deform specimens at strain rates of the order 10^5 s^{-1} . However, Charpy impact data is useful for qualitatively determining the effect of strain rate on material behaviour when compared to quasi-static data.

The Charpy impact data obtained from [84] consisted of measured σ_{UTS} values from a limited number of tests performed in the longitudinal tube direction on as-received and irradiated Zr-2.5Nb pressure tube material at temperatures ranging from 20 to 300°C. The irradiated pressure tube material had been removed from operating CANDU reactors and had fluences ranging from approximately 1.3×10^{25} to $7.5 \times 10^{25} \text{ n/m}^2$. Small beam

specimens were manufactured from the pressure tubes along the longitudinal direction and tested with an instrumented impactor so that the applied load could be measured and used to calculate the σ_{UTS} for each test. Figure 1.22 shows the σ_{UTS} as a function of temperature for those tests. Included in Figure 1.22 are results obtained by Ibrahim [81] at quasi-static strain rates on irradiated longitudinal specimens. It is seen that the σ_{UTS} values measured at the higher strain rate are significantly greater than those measured under quasi-static loading conditions. These results indicate that the effect of increasing the strain rate is to increase the strength of irradiated Zr-2.5Nb pressure tube material for temperatures between room temperature and 300°C.

1.2.6 Summary of the Tensile Mechanical Behaviour

Although there is a limited amount of data pertaining to the deformation behaviour of Zr-2.5Nb pressure tube material under quasi-static tensile loading conditions, it appears consistent and suggests that pressure tube material is stronger in the transverse direction, independent of temperature and irradiation level. The unique stress-strain behaviour exhibited, as a function loading direction, is due to the unique crystallographic texture and the limited number of deformation systems in α -zirconium. That is, Zr-2.5Nb pressure tube material will deform primarily by prismatic slip when loaded in the longitudinal direction and twinning in the transverse direction. As a result, a significant amount of work hardening occurs in the longitudinal direction. This behaviour is consistent for as-received and irradiated material over the temperatures ranging from room temperature to approximately 300°C. When irradiated pressure tube

material is loaded quasi-statically in tension in the transverse direction, at temperatures in the range of room temperature to approximately 100°C, the σ_{UTS} occurs at very low plastic strains close to the σ_y , indicating a low work hardening capacity. The material then strain-softens rapidly until failure. For temperatures in the range of approximately 100 to 300°C, strain localisation occurs as a result of co-operative twinning causing the stress-strain behaviour to vary for different tubes. However, essentially no work hardening occurs and the values of σ_y and σ_{UTS} remain close.

The overall effect of irradiation is to increase the strength and decrease the ductility for both orientations and for temperatures ranging from room temperature to 300°C. A large increase in strength is observed for irradiation fluences up to approximately $1 \times 10^{25} \text{ n/m}^2$ followed by a small, but measurable increase to a fluence of $17 \times 10^{25} \text{ n/m}^2$, corresponding to an initial rapid increase in the density of $\langle a \rangle$ -type dislocations and a slow, but steady increase in $\langle c \rangle$ -type dislocations, respectively.

The effect of increasing the temperature is to decrease the strength of as-received and irradiated material in both the longitudinal and transverse directions and increase the ductility. For temperatures increasing above approximately 300°C, the effect of irradiation strengthening is annealed out.

The effect of increasing the strain rate from quasi-static conditions to approximately 80 s^{-1} is to increase the strength of as-received and irradiated pressure tube material for temperatures ranging from room temperature to 300°C.

1.3 Material Behaviour at High Rates of Strain

To develop an accurate description of the mechanical response for a material, an adequate understanding of the rate dependency is required [7]. Data of this nature has been obtained for a large number of materials and it was observed that most metals and alloys exhibit an increase in strength with increasing rates of strain, with higher rate sensitivity at higher rates [1 - 9]. This behaviour is more apparent in pure metals and increases in alloying content usually result in a reduced strain rate sensitivity [7]. Although most data concerning rate sensitivity has been obtained for face-centred-cubic (fcc) metals, those with bcc, or hcp structure tend to exhibit the greatest strain rate sensitivity [85, 86]. The rate sensitivity of fcc metals has received greater attention since the response of bcc and hcp metals are more complex. In order to understand the difference in rate sensitivity for various crystallographic structures, the associated dislocation mechanisms must be considered [7, 85, 86].

Another factor for materials undergoing deformation at high strain rates is adiabatic heating [87]. When metals and alloys are deformed to large strains under high strain rate conditions, there may be insufficient time for the heat generated during plastic deformation to diffuse away from the deforming area. The reduction in strength due to the local increase in temperature, known as thermal softening, can be greater than the increase in flow stress due to strain, or strain rate hardening. In this situation, further plastic deformation leads to further local thermal softening in an accelerated fashion. As a result, plastic flow can become completely restricted to a localised band, known as an adiabatic shear band [87].

1.3.1 General Description of Mechanical Behaviour

It is common to show the yield stress or flow stress, at a given value of plastic strain, as a function of logarithm of the strain rate when presenting the strain rate sensitivity of a material [87]. Based on a large compilation of data presented in this manner, Campbell [88] identified three distinct regions of strain rate response, as shown in Figure 1.23. The regions corresponded to conditions of rate-independent flow (Region I), linear dependence of flow stress on log-strain rate (Region II), and increased dependence on log-strain rate (Region III). Region I corresponds to conditions of low strain rate and high temperature, where flow stress is approximately constant and independent of strain rate and temperature. In this region, flow stress is controlled by long-range friction stress resulting from dislocation barriers such as large precipitates. Here, thermal vibrations in the lattice can not assist in overcoming these barriers and, therefore, the friction stress is considered athermal. The athermal friction stress increases for increasing alloy content and predominates as the controlling mechanism to flow in highly alloyed metals, resulting in less rate sensitivity [7].

In Region II the flow stress depends linearly on the logarithm of strain rate and corresponds to conditions of higher strain rates and lower temperatures than Region I. Here, short-range barriers to flow become the controlling mechanism. Thermal vibrations can assist in overcoming barriers, such as dislocation interaction, and therefore, flow becomes sensitive to temperature and strain rate. The rate of plastic strain ($\dot{\epsilon}_p$) is related to the absolute temperature (T) by an Arrhenius function of the form,

$$\dot{\varepsilon}_p = \dot{\varepsilon}_0 e^{(-\Delta G/kT)} \quad (1.3)$$

where ΔG is the free energy of activation, and $\dot{\varepsilon}_0$ is the limiting strain rate. Generally, $\dot{\varepsilon}_0$ is taken to be independent of $\dot{\varepsilon}_p$ and T , while ΔG is expressed as a unique function of stress [7]. Assuming that a single thermally activated mechanism controls flow, ΔG may be expressed as a linear function of stress as,

$$\Delta G = \Delta G_0 - V(\tau - \tau_a) \quad (1.4)$$

where ΔG_0 is the free energy of activation in the absence of stress, V is the activation volume, τ is the total applied stress and τ_a is the athermal component of stress associated with Region I. Combining equations 1.3 and 1.4 provides the relationship,

$$\tau = \tau_a + \frac{\Delta G_0}{V} + \frac{kT}{V} \ln \left(\frac{\dot{\varepsilon}_p}{\dot{\varepsilon}_0} \right) \quad (1.5)$$

which corresponds to very high strain rates (Region III). Taking V to be independent of stress and assuming $\sigma = \sqrt{3} \tau$, the logarithmic strain rate sensitivity parameter (λ) is defined [7] as,

$$\lambda = \left(\frac{\partial \sigma}{\partial \log(\dot{\epsilon}_p)} \right)_{T, \epsilon_p} = \frac{4kT}{V} \quad (1.6)$$

For Region I, λ approaches zero. In Region II, dislocations require thermal energy to overcome obstacles and λ attains a positive value that is generally constant. In Region III, λ increases rapidly above values obtained in Region II for increasing strain rate. The rate controlling mechanisms [89, 90] causing the observed behaviour of Region III are currently under debate [91]. For the range of strain rates and temperatures considered in the current investigation, the behaviour depicted in Regions I and II is expected.

The accuracy of equations 1.5 and 1.6 in predicting the mechanical response of metals and alloys appears to depend on material structure [7, 87]. Those with an fcc structure generally show a reasonable correlation, while metals and alloys with a bcc structure tend to follow the predicted response less closely. Metals and alloys with a hcp structure typically have a more complicated rate-dependence, showing similarities with both fcc and bcc metals, depending on the specific metal or alloy.

For 1100-0 aluminum (fcc), Lindholm [92] showed that below a strain rate of approximately $1 \times 10^{-3} \text{ s}^{-1}$ the flow stress was athermal, while above this strain rate a constant value of λ was obtained, as predicted by equations 1.5 and 1.6. A similar conclusion was reported for fcc copper and lead [7]. In [93] it was suggested that the strain rate at which athermal behaviour occurs varies with the material being tested, and

for 6061-T6 and 7075-T6 aluminum (fcc) athermal behaviour was observed for strain rates up to approximately $1 \times 10^3 \text{ s}^{-1}$.

Mild steel (bcc) tested in conditions associated with Region II had constant values of λ for a given temperature, which agrees with predicted behaviour. However, data from tests performed at different temperatures indicated that λ decreased for increasing temperature [88], which does not agree with equation 1.6. Two other bcc metals, high purity iron [94] and niobium [95], exhibited non-linear strengthening as a function of log-strain rate over the entire thermally activated region (Region II). This behaviour was also reported [96] for alloy steels tested in tension at strain rates over approximately 10 s^{-1} . Below strain rates of 10 s^{-1} and at room temperature, however, the flow stress was rate independent, which agrees with equation 1.6.

Figure 1.24 shows the flow stress at 5% plastic strain for α -titanium (hcp) as a function of log-strain rate and temperatures ranging from 15 to -200°C [97]. From Figure 1.24 it is apparent that a bi-linear dependence of flow stress on log-strain rate exists with a higher sensitivity over the high strain rate region. Different values of λ were obtained for a given temperature above and below a strain rate of approximately 100 s^{-1} . Over the quasi-static strain rate regime λ decreased for decreasing temperature, which agrees with behaviour predicted by equation 1.6. Conversely, in the high strain rate region, λ appeared to be independent of temperature.

Investigations into the high strain rate properties of the hcp titanium alloy, Ti-6Al-4V, [56, 98 - 100] and high-purity titanium [58, 61] have demonstrated that strength increases significantly for increasing strain rates. The rate sensitivity, for strain rates achieved using the split Hopkinson bar (maximum $\sim 4000 \text{ s}^{-1}$), was attributed to

thermally activated mechanisms [56, 58, 61]. It was further concluded that because of its high yield stress, low heat capacity and thermal conductivity, adiabatic heating had a significant effect on the stress-strain behaviour at strain rates greater than 100 s^{-1} [56, 99].

1.3.2 Behaviour of Zirconium and Zirconium Alloys

As described in Section 1.2.5, there is essentially no data available that can be used to characterise the high strain rate behaviour of Zr-2.5Nb pressure tube material. Recent studies performed on pure zirconium [59, 60] and Zircaloy[®] [101], however, provide insight into the effect of high strain rate loading on the mechanical behaviour of zirconium. Since pressure tube material is so highly textured and behaves as a pseudo-monocrystal [74], these studies should provide qualitative information regarding the high strain rate behaviour of Zr-2.5Nb pressure tube material. Tensile tests performed on Zircaloy-4[®] specimen [101] at strain rates of 1×10^{-3} , 450, and 1500 s^{-1} , and temperatures of 20 and 300°C indicated that the σ_{UTS} increased and the ductility decreased for increasing strain rate. As expected, the opposite effect was observed for increasing temperature. That is, the σ_{UTS} decreased and the reduction in area increased for increasing temperature. This behaviour was consistent for four material conditions considered: stress-free annealed, as-rolled, fully annealed and hydrogen embrittled ($\sim 200 \text{ ppm}$).

Recently, the effects of temperature, strain rate, interstitial content, and texture on the compressive behaviour of high-purity and low-purity zirconium were investigated [60]. Tests were performed at temperatures ranging from -200°C to approximately

700°C and strain rates ranging from $1 \times 10^{-3} \text{ s}^{-1}$ to approximately 3000 s^{-1} . The high-purity material contained approximately 250 ppm of impurities and the low-purity material contained approximately 1000 ppm of impurities. Both materials were cold-worked to yield a strong basal texture and tested in directions parallel to the *c*-axis (transverse specimen) and perpendicular to the *c*-axis (in-plane specimen). The effect of increasing the strain rate was to increase the strength for both specimen orientations and for all temperatures. For all of the test temperatures and strain rates considered, the transverse specimen were considerably stronger than the in-plane specimen. It was concluded that the transverse specimen were stronger since the direction of loading, with respect to the *c*-axis, was preferentially oriented for deformation twinning [60, 61]. Based on the stress-strain responses of the transverse specimen, there was a transition in the work hardening behaviour as a function of temperature and strain rate [59, 60]. The distinct work hardening behaviour was shown to be caused by increased amounts of twinning [60, 61]. That is, as the temperature was decreased, or the strain rate increased, the propensity for twinning increased. Increases in the interstitial content corresponded to a reduced incidence of twinning in specimens tested under quasi-static conditions [59 - 61]. It was speculated that similar behaviour would occur for specimens tested at high strain rates [60, 61].

1.3.3 Adiabatic Heating

When a metal or alloy is plastically deformed, the majority of the plastic work is converted into heat [87]. It is generally accepted that in excess of 85% of the energy

associated with plastic deformation is dissipated as heat, with the remaining energy being stored in the specimen in the form of new defect structures. The increase in temperature (ΔT) resulting from plastic work during adiabatic deformation is [87],

$$\Delta T = \frac{\beta}{\rho c_p} \int_{\varepsilon_p=0}^{\varepsilon_p=\varepsilon'} \sigma d\varepsilon_p \quad (1.7)$$

where β is the fraction of plastic work converted to heat, ρ is the density of the material, c_p is the specific heat of the material, σ is the stress, and ε_p is the plastic strain. The fraction of the plastic work that is converted to heat is generally taken to be in the range of 85 to 95% [87]. In a recent study involving the high strain rate testing ($\sim 2500 \text{ s}^{-1}$) of Remco iron at room temperature on a tensile split Hopkinson bar [102], temperatures in the fracture process region were measured and ranged from 250 to 350°C.

1.4 High Strain Rate Testing of Materials

As the research associated with this thesis is concerned with the tensile behaviour of Zr-2.5Nb pressure tube material, the split Hopkinson pressure bar configuration for testing materials in tension will be reviewed in detail. First, however, the experimental techniques used to test materials at high rates of strain and the theoretical basis of the split Hopkinson technique will be discussed.

1.4.1 Experimental Techniques

John Hopkinson in 1872 [103, 104] was the first to experimentally demonstrate that metals could withstand higher levels of stress when loaded dynamically in tension. Following this work, his son Bertram Hopkinson [105, 106] went on to develop an apparatus that was used to determine the profile and evolution of pulses that would travel down long rods when loaded by impact at one end. The portion of the apparatus that was used to transmit the force pulse, or elastic stress-wave, consisted of a long slender rod, which became known as a Hopkinson bar. In 1949, Kolsky [107] published a paper outlining the use of two such bars in a manner which enabled him to determine the dynamic behaviour of metallic test specimen loaded in compression. The experimental method presented by Kolsky [107], known as the Kolsky or split Hopkinson pressure bar technique, remains the most accepted method for determining the dynamic stress-strain behaviour of materials [8, 9]. Since the introduction of the compression Hopkinson apparatus by Kolsky [107], other configurations of the split Hopkinson pressure bar were developed for testing in tension [1, 108], in torsion [109], and under biaxial loading [110]. Today there are a number of variations for each configuration and a review by Field et al. [9] identified more than 1000 papers in the open literature on split Hopkinson pressure bar testing of metals, polymers, ceramics, and rocks. Alternative experimental techniques for determining the behaviour of materials at high strain rates include the expanding ring, or cylinder test, the Taylor cylinder impact test, long-rod wave propagation tests, drop forge tests, and dynamic bend tests [4, 8, 111].

In the expanding ring / cylinder test arrangement, a state of uniaxial plane stress, or plane strain is achieved through the sudden symmetrical radial acceleration of a ring or cylinder, respectively. The ring or cylinder, once dynamically loaded with an initial impulse, expands radially outwards at some initial radial velocity. A dynamic stress-strain relationship is then obtained by considering the deceleration and strain of the outer wall as a function of time. Based on this idea, Clark and Duwez [112] introduced a technique in which a piston driven system was utilised to load thin-walled steel cylinders. In their experiments [112] the cylinders were filled with mercury and a piston was driven into the cylinder at a constant velocity, thereby dynamically pressurizing the cylinder. Other means of dynamic loading expanding rings and cylinders to test materials in this manner have been achieved through the use of electric discharges, electromagnetic forces, and explosive charges [4, 8, 111]. Irrespective of the loading method, this technique is complicated by the transient nature of the initial loading history and strain rate during deceleration. These difficulties and those associated with obtaining accurate experimental measurements have limited this technique in characterising material behaviour [4, 8, 111].

The Taylor cylinder test method, introduced by G.I. Taylor in 1948 [113], involves firing a solid cylinder of material at a massive and rigid target. Based on measurements of only the initial and final dimensions of the cylinder along with the velocity at impact, the dynamic yield stress can be calculated. When introduced, Taylor used a simple analysis assuming rigid, perfectly plastic material behaviour, one-dimensional wave propagation concepts, and a frictionless boundary condition at the cylinder/target interface to derive the dynamic yield stress formula [113]. The original

test technique and analysis method have been critically examined by a number of investigators resulting in modifications that take better account of material behaviour, boundary conditions and the three-dimensional, transient nature of the test method [4, 8]. The Taylor cylinder test, however, remains a simple method of determining some information concerning the dynamic properties of materials and the prevalent test that is numerically modelled to validate material constitutive relations [111].

In long-rod wave propagation tests, a rod is impacted at one end at a velocity high enough to cause plastic deformation. The rod is instrumented so the strain-time histories can be recorded at a number of locations along the bar and used to monitor the velocity of the resulting elastic-plastic wave. Although this test method does provide information on the dynamic properties of materials, it is not commonly used to determine the constitutive behaviour [4, 8]. In order to perform the analysis, the material behaviour must be known. However, it is the material behaviour that is desired in the first place. Therefore, determination of the constitutive behaviour is indirect and requires an iterative process.

1.4.2 Theoretical Basis of the Split Hopkinson Bar Technique

Independent of the Kolsky, or split Hopkinson pressure bar configuration, the overall technique as presented by Kolsky [107] involves placing a short specimen between an input and an output bar [4, 8], as shown schematically in Figure 1.25. The specimen is loaded by an elastic stress wave (incident wave) that is generated in the input bar. This incident wave propagates down the input bar and loads the specimen. Part of the input wave is transmitted through the specimen and propagates along the output bar

(transmitted wave), and part is reflected back into the input bar (reflected wave). Assuming the strain is uniform along the length of the specimen, the strain, stress, and strain rate of the specimen can be calculated from the equations describing one-dimensional, elastic-wave propagation in rods [114, 115] and the time-histories of strains recorded in the bars [107].

Figure 1.26 is a schematic representation of the specimen and the ends of the Hopkinson bars, showing the incident (ϵ_i), reflected (ϵ_r) and transmitted (ϵ_t) strain pulses. If the subscripts 1 and 2 are used in Figure 1.26 to represent the end of the input and output bars, respectively, the displacements at the ends of the specimen, as a function of time (t), can be expressed as,

$$u_1 = c_0 \int_0^t \epsilon_1 dt \quad (1.8)$$

and

$$u_2 = c_0 \int_0^t \epsilon_2 dt \quad (1.9)$$

where ϵ_1 and ϵ_2 are the instantaneous strains in ends 1 and 2 of the Hopkinson bars, and c_0 is the longitudinal elastic-wave velocity in the Hopkinson bars. Now, in terms of the incident (ϵ_i), reflected (ϵ_r), and transmitted strains (ϵ_t), equations 1.8 and 1.9 become,

$$u_1 = c_0 \int_0^t (\varepsilon_i - \varepsilon_r) dt \quad (1.10)$$

and

$$u_2 = c_0 \int_0^t \varepsilon_t dt \quad (1.11)$$

where the stresses and strains are assumed to be positive in tension. The engineering strain along the specimen gauge length (ε_s), can then be expressed as,

$$\varepsilon_s = \frac{u_1 - u_2}{L} \quad (1.12)$$

or, in terms of the strain pulses,

$$\varepsilon_s = \frac{c_0}{L} \int_0^t (\varepsilon_i - \varepsilon_r - \varepsilon_t) dt \quad (1.13)$$

where L is the gauge length of the specimen. Now, since

$$\dot{\varepsilon} = \frac{d\varepsilon}{dt} \quad (1.14)$$

the strain rate of the specimen ($\dot{\epsilon}_s$) can be calculated from,

$$\dot{\epsilon}_s = \frac{c_0}{L}(\epsilon_i - \epsilon_r - \epsilon_t) \quad (1.15)$$

This equation shows that the strain rate as a function of time depends directly on the elastic wave velocity and inversely on the specimen's gauge length. The forces at the ends of the specimen (P_1 and P_2) can be calculated by,

$$P_1 = EA(\epsilon_i + \epsilon_r) \quad (1.16)$$

and

$$P_2 = EA\epsilon_t \quad (1.17)$$

where E and A are the Young's modulus and cross-sectional area of the Hopkinson bars, respectively. The average force acting on the specimen (P_{ave}), calculated as the average of equations 1.16 and 1.17, is

$$P_{ave} = \frac{EA}{2}(\epsilon_i + \epsilon_r + \epsilon_t) \quad (1.18)$$

and for a specimen of cross-sectional area A_s , the stress history of the specimen (σ_s) with time can be calculated from,

$$\sigma_s = \frac{EA}{2A_s}(\varepsilon_i + \varepsilon_r + \varepsilon_t) \quad (1.19)$$

This equation shows that the stress in the specimen is linearly proportional to the sum of the amplitudes of the incident, reflected and transmitted waves. By integrating the strain rate data to obtain the strain history and combining with the stress history, the stress-strain behaviour of the material is obtained.

Davies [116] critically examined the one-dimensional assumption adopted by Kolsky [107] and used the equations of motion describing sinusoidal wave propagation in an infinite cylindrical bar [114, 115] to describe the three-dimensional effects of dispersion and inertia in the bars. The analysis [116] identified that the difference between the strain at the center-line and outer surface of the bar would be less than 5% for a bar with a radius to pulse length less than 0.1. Therefore, the assumption of one-dimensional, elastic-wave propagation in rods is valid when tests involve split Hopkinson bars of small diameter and pulses of long duration.

Davies and Hunter, [117] also examined the assumption of uniform deformation used in the development of the Hopkinson equations by Kolsky [107]. It was noted [117] that an initial transition time is associated with the development of a state of uniform strain in the specimen. During this time a non-uniform state of deformation exists, due to longitudinal and radial inertial effects, and the Hopkinson equations are not valid. They concluded that

after π reverberations of the stress pulse along the specimen, equilibrium is nearly achieved and inertial effects would not greatly influence the results. It was suggested that the time required to reach uniform deformation ($t_{uniform}$) is [117],

$$t_{uniform}^2 > \frac{\pi^2 \rho_s l_s^2}{\partial \sigma / \partial \epsilon} \quad (1.20)$$

where ρ_s and l_s are the specimen density and length, respectively, and $\partial \sigma / \partial \epsilon$ is the work hardening rate of the true-stress-true-strain curve for the material. From equation 1.20 it is seen that by decreasing the specimen length, the time required to reach a state of equilibrium is decreased.

Recent experimental [118] and numerical [119, 120] investigations examined the effect of specimen geometry on the state of deformation for specimens tested in tension using the split Hopkinson bar apparatus. In these investigations it was demonstrated that geometry can have a significant effect on the calculated value of stress for a tensile specimen. It was concluded that the specimen length (l_s) to diameter (d_s) ratio should be greater than approximately 1.6 to ensure the calculated value of stress is accurate. For values of $l_s / d_s < 1.6$, the stress state within the gauge length is not uniform and the stress obtained from the Hopkinson equations will be larger than the material's actual stress, for a given value of strain.

Since the Hopkinson equations assume a state of uniform strain along the specimen, data obtained beyond the point of strain localisation cannot be used directly to calculate the

true-stress as a function of true-strain response of the material. The reduction in area as a function of engineering strain must be determined and used in conjunction with the measured engineering stress and strain to determine the materials true properties [111].

1.4.3 Tensile Configuration of the Split Hopkinson Bar Technique

Since the introduction of the first split Hopkinson bar for testing in tension [108], a number of other configurations have been introduced [4, 8, 111]. The primary differences in the different types of tensile bars are the methods in which the incident pulse is generated, the specimen geometry, and the methods of attaching the specimen to the input and output bars. The various tensile split Hopkinson bar configurations can be separated into three basic types. The first type utilises a compressive pulse and a novel specimen design [1], as shown in Figure 1.27. The input (incident) bar is solid and the output (transmitted) bar is a hollow tube with the same cross-sectional area as the input bar (Figure 1.27a). As shown in Figure 1.27b, the specimen consists of four gauge lengths and fits inside the output bar with the input bar fitting inside the specimen. The advantage of this configuration is that the tests can be done in the same manner as a compression test with little modification to an existing compression bar. The main disadvantage is that the specimens are difficult to manufacture.

The second type of tensile split Hopkinson bar configuration utilises a specimen that is rigidly attached to the ends of the input and output bars [108, 118, 121] and a tensile pulse that is applied directly to the input bar. The specimen generally has ends that thread into holes in the ends of the bars [108, 121], but in other designs the specimen is glued onto the bars to eliminate the use of mechanical joints [118]. For this type of configuration the tensile

load is applied in a number of different ways, as shown schematically in Figure 1.28. In Figure 1.28 (a), a mass is impacted directly onto an anvil that is attached to the end of the input bar. In Figure 1.28 (b), a hollow loading bar impacts a long anvil that is attached to the input bar causing the formation of a compression pulse. The compression pulse travels to the end of the anvil and reflects back and travels as a tensile wave down the input bar. In Figure 1.28 (c), the detonation of an explosive charge in contact with the anvil is used to generate the tensile pulse in the input bar. In Figure 1.28 (d), the input bar is clamped mid-way down and a static tensile load is applied to the end. To generate an incident tensile pulse, a fracture pin holding the clamp is broken thereby quickly releasing the stored tensile load. An advantage of this type of bar is that a relatively slow rise time (~ 30 to $40 \mu\text{s}$) is associated with the incident pulse, which minimises inertial effects. Also, because of the method of creating the incident pulse, there is little noise associated with the incident wave. A disadvantage, however, is that extra effort is required to align the bar and clamping mechanism [118]. For tensile bars as depicted in Figure 1.28 (b) and (d), a constant strain rate of considerable duration can be achieved. For the bars depicted in Figure 1.28 (a) and (c), it is difficult to generate an incident pulse of constant magnitude.

The third type of tensile bar configuration also uses a threaded type of specimen, as shown in Figure 1.29 [4]. The specimen design includes a solid hollow shoulder that fits over the specimen with the same outside diameter as the input and output bars, but is not attached to the bars. That is, the hollow shoulder is held in place by tightening the specimen into the bars, thereby squeezing the shoulder between the bars. The test is conducted exactly as a compression test, but relies on the reflected compression pulse to load the specimen in tension [4]. The incident compression wave travels through the shoulder into the output bar

without loading the specimen. When the incident compression wave reaches the end of the output bar, it reflects back as a tensile wave. When the tensile wave reaches the specimen, the hollow shoulder is unable to support any of the tensile load since it is not attached to the bars. Part of the tensile pulse, therefore, travels through the specimen as the transmitted pulse thereby loading it in tension, and the remainder is reflected back as the reflected pulse. The major shortcoming of this method is that it is practically impossible to transmit all of the initial compression pulse through the specimen-shoulder configuration without having spurious wave reflections [4].

1.5 Constitutive Relations

The mathematical formulation describing the macroscopic response of a material to applied forces is known as a constitutive relation, or equation [122]. In the most general form, a constitutive equation relates the stress of a material to the strain, strain rate, temperature, and loading history experienced by the material. It is not possible, however, for a single relationship to completely describe the deformation behaviour of a material for all combinations of loading conditions and histories [122]. Depending on the intended use, availability of experimental data, and the desired accuracy of the description, a decision must be made as to the amount of effort to be invested. Typically, the effort required to describe the material behaviour is proportional to the degree of accuracy with which the governing physical mechanisms are modelled [8].

The availability of current high-speed computers allows for the implementation of more complicated, physically-based constitutive relations into codes for solving problems

involving deformation of materials at high strain rates. The ability to accurately describe material behaviour for a wide range of loading conditions and to predict the mechanical response outside of the measured conditions is growing in demand. Despite this, most of the constitutive relations implemented into computer codes describe only the effects of strain hardening, and take some account of thermal softening, but few relations presently account for strain rate effects [8].

Three of the most widely [8, 123, 124] used relations that do account for strain rate effects are the Johnson-Cook [125, 126], the Zerilli-Armstrong [127 - 130], and the mechanical threshold stress (MTS) [131] constitutive relations. The Johnson-Cook and the Zerilli-Armstrong constitutive relations are relatively simple, one dimensional models that account for the effects of strain, strain rate, and thermal softening on flow stress and utilise a von Mises yield criterion [8, 123, 124]. They both describe the material hardening behaviour based on the well known power-law function introduced by Ludwick [132]. That is,

$$\sigma = \sigma_0 + K \cdot \epsilon^n \quad (1.21)$$

where σ_0 , K , and n , are material constants.

The Johnson-Cook constitutive relation is an empirical relation that is relatively simple to calibrate for a given material. That is, very few stress-strain curves covering the loading conditions are required to determine the parameters. It is relatively easy to implement into computer codes, inexpensive to use and produces reasonably accurate

predictions for a range of materials if loading conditions do not exceed those used to determine the material parameters.

The Zerilli-Armstrong constitutive relations are semi-empirical models based on the concept of dislocation dynamics. Separate relations were developed to describe the deformation behaviour of copper, which is fcc, and iron, which is bcc. A subsequent relation was introduced for hcp materials which generally exhibit behaviour between that of bcc and fcc materials. Overall, the Zerilli-Armstrong relations tend to give a slightly better correlation to experimental results when compared to the Johnson-Cook relation [8, 124]. The Zerilli-Armstrong relation requires slightly more data to calibrate than the Johnson-Cook relation, but is still relatively easy to implement into computer codes and inexpensive to use. Similar to the Johnson-Cook relation, however, it is not very accurate when used outside of the range of loading conditions that it was calibrated for, or for large strain calculations [124].

The MTS relation is a more advanced and physically-based model that relates the macroscopic mechanical response of a material to microscopic properties [131]. It accounts for the evolution of sub-structures as a function of plastic deformation and loading history. It generally is more accurate than both the Johnson-Cook and Zerilli-Armstrong relations, and once it is calibrated for a particular material it can be used to accurately predict deformation outside the range of calibration loading conditions [124]. This has been attributed to the form used to represent the work hardening behaviour in the MTS relation. Rather than use a power-law form [132], which yields a continuously increasing stress for increases in strain, Follansbee and Kocks [131] selected a function that “saturates” to a maximum stress level for increasing strain. This concept was based on the observation

[133] that at large strains most metallic materials tend to approach a finite “saturation stress”, or approach a constant, but small hardening rate at large strains [134, 135]. However, it should be noted that the form of the equation used to describe the work hardening behaviour is not strictly physically-based and, therefore, the MTS relation is also semi-empirical. The main disadvantages of the MTS relation are that it requires an extensive experimental data base to determine the material parameters, it is more difficult to incorporate into computer codes and more computationally expensive.

Another classification of constitutive relations that are used in problems involving plastic deformation at high strain rates are unified viscoplastic relations. These relations are generally formulated in three-dimensions and are based on the decomposition of strain rate into elastic and plastic components, where the plastic strain rate is time dependent and can represent all possible non-elastic material processes. Unified viscoplastic relations can be used to calculate loading history effects and problems involving directional hardening. In [136], a comparison between calculations performed using the Johnson-Cook relation and a unified viscoplastic relation, the Bodner-Partom model [137], was conducted and showed good agreement. The main disadvantages of viscoplastic relations are that they are more difficult to incorporate into codes and require iterative solutions making them computationally expensive. Consequently, they have seen limited use in codes utilised for high strain rate material modelling [8].

The following two sections provide further details of the Johnson-Cook and Zerilli-Armstrong constitutive relations as they are used in the current research to describe the deformation behaviour of as-received and irradiated Zr-2.5Nb pressure tube material.

1.5.1 Johnson-Cook Constitutive Relation

The Johnson-Cook constitutive relation [125, 126] is a one-dimensional, empirically-based relation that describes the variation of flow stress ($\bar{\sigma}$) with respect to strain, strain-rate and temperature as,

$$\bar{\sigma} = (A + B\bar{\epsilon}_p^n)(1 + C \ln \dot{\epsilon}^*)(1 - T_h^m) \quad (1.22)$$

where $\bar{\epsilon}_p$ is the effective plastic strain, $\dot{\epsilon}^*$ is a non-dimensional, effective plastic strain rate, and T_h is a form of homologous temperature. The parameters A , B , C , m , n are material constants obtained from fitting experimental data and are defined as follows: A is the yield strength under static loading conditions, B and n are the work hardening coefficient and exponent, respectively, C is the strain rate coefficient, and m is the thermal softening exponent.

The non-dimensional strain rate is defined as [125],

$$\dot{\epsilon}^* = \frac{\dot{\epsilon}}{\dot{\epsilon}_0} \quad (1.23)$$

where $\dot{\epsilon}_0$ is a reference strain rate, typically taken to be 1 s^{-1} , and the homologous temperature is defined as [125],

$$T_h = \frac{(T - T_{room})}{(T_{melt} - T_{room})} \quad (1.24)$$

where T is the absolute temperature of the specimen, T_{room} represents room temperature, and T_{melt} is the melting temperature. It should be noted that all of the temperatures are in units of degrees K.

The Johnson-Cook relation is essentially a modified form of the over-stress type of models introduced by Malvern [138], Perzyna [139] and Campbell [140]. In these models, the plastic strain rate is not limited to a finite value and therefore, the strength increases linearly with respect to logarithmic strain rate. It should also be noted that effects of non-proportional loading history on flow stress are not considered in this model.

The Johnson-Cook relation considers the effects of strain, strain rate, and temperature in an uncoupled manner, as seen in equation 1.22. The first term describes the flow stress as a power-law of strain. The second term represents the logarithmic effect of strain rate, and the third term expresses the effect of temperature through thermal softening. Since these terms are uncoupled it implies that the strain rate sensitivity is independent of temperature, which is typically not true for metals [8].

Holmquist and Johnson [141] introduced a modified version of the Johnson-Cook relation based on evidence that the influence of strain rate on material strength is not a linear function of the natural log. They suggested that it is better represented as an exponential function. The flow stress in the modified Johnson-Cook relation [141] is represented as,

$$\bar{\sigma} = (A + B\bar{\epsilon}_p^n)(\dot{\epsilon}^*)^m(1 - T_h^m) \quad (1.25)$$

where the parameters are defined as above.

1.5.2 Zerilli-Armstrong Constitutive Relations

Zerilli and Armstrong [127] introduced two dislocation-mechanics based constitutive relations that describe the von Mises flow stress as a function of strain, strain rate, temperature and grain size in a coupled manner. Based on the concept of thermally activated motion of dislocations and the nature of dislocation interactions, one relation was developed to describe the deformation of fcc copper and the other for bcc iron. Although the constitutive relations introduced by Zerilli and Armstrong [127] were developed for copper and iron, it was intended that they would be applicable for a wide range of fcc and bcc materials since they were developed from physical principles inherent to each crystallographic structure.

For bcc metals, the motion of dislocations is governed predominantly by the interaction produced by the overall lattice potential (Peierls-Nabarro stress), which leads to minimal strain hardening. For fcc metals, the predominant interaction is at dislocation intersections, which leads to substantial strain hardening. The equation for bcc metals is,

$$\bar{\sigma} = \sigma_a + B e^{-\beta T} + B_0 \bar{\epsilon}^n \quad (1.26)$$

and for fcc metals it is,

$$\bar{\sigma} = \sigma_a + B_0 \bar{\varepsilon}^{1/2} e^{-\alpha T} \quad (1.27)$$

where $\bar{\varepsilon}$ is the equivalent plastic strain, and T is the absolute temperature. The parameter σ_a represents the athermal component of flow stress and is determined mainly from the effects of solutes and grain boundaries. It is expressed as,

$$\sigma_a = \sigma_G + k d^{-1/2} \quad (1.28)$$

where, σ_G represents the influence of solute and the original dislocation density on the effective stress. The second term, known as the Hall-Petch term, calculates the stress associated with the plastic flow between grain boundaries at low temperatures. The variable k represents the microstructural stress intensity and the variable d is the average grain diameter.

The terms α and β are expressed as,

$$\alpha = \alpha_0 - \alpha_1 \ln \dot{\varepsilon} \quad (1.29)$$

$$\beta = \beta_0 - \beta_1 \ln \dot{\varepsilon} \quad (1.30)$$

where $\dot{\varepsilon}$ is the equivalent strain rate, and the variables $B, B_0, \beta_0, \beta_1, \alpha, \alpha_0, \sigma_G, k, d$ and n are material constants that are determined from experimental data.

Zerilli and Armstrong [127] implemented both constitutive relations for fcc and bcc metals into the computer program EPIC-2 and simulated the Taylor cylinder impact tests conducted by Johnson and Cook [125, 126]. The results from their simulations were compared to simulations conducted in [126] and it was concluded that the results from the Zerilli-Armstrong relation more closely matched the true deformation behaviour for the copper and iron. In a similar study by Holmquist and Johnson [141] the same conclusion was made.

Zerilli and Armstrong [128] stated that their original relation for bcc metals, expressed in equation 1.26, was incomplete since it did not account for the effect of twinning, which is important in bcc materials. An improved version for the bcc relation was introduced [128], which incorporated a simple deformation twinning model. It was suggested that plastic flow by slip could be replaced by twinning at high strain rates in the presence of large grains. To account for this effect, they modified the Hall-Petch term in equation 128 to be,

$$\sigma_a = \sigma_G + kd^{-1/2} [(N_T + 1)^{1/2} - 1] \quad (1.31)$$

where N_T represents the average number of twins per grain. The improved bcc relation was implemented into EPIC-2 and the Taylor cylinder impact tests of Johnson and Cook [125, 126] were again simulated. The results indicated an overall increase in accuracy when modelling iron.

A basic assumption of the Zerilli-Armstrong bcc relation, equation 1.26, is that the work hardening rate is independent of both strain rate and temperature. In 1991,

Goldthorpe [142] confirmed this independence with respect to strain rate, but demonstrated that the work hardening rate is dependent on temperature. The effect of temperature on work hardening rate was negligible for small strains and small increases in temperature, but significant at large strains and with increases in temperature [142] due to adiabatic heating. A modification to the original Zerilli-Armstrong bcc relation was introduced of the form,

$$\bar{\sigma} = \sigma_a + B e^{-\beta \dot{\epsilon}} + (B_0 \bar{\epsilon}^n + C) \frac{\mu_T}{\mu_{293}} \quad (1.32)$$

where C is a constant and $\frac{\mu_T}{\mu_{293}}$ is the ratio of the shear modulus at the test temperature with respect to the shear modulus at room temperature (taken to be 293 K in [142]). Shear modulus is assumed to be a function of temperature and represented by a second order polynomial [142]. The modified version of the bcc relation was used in numerical simulations of split Hopkinson bar experiments and explosively formed projectiles (EFP's), and it was concluded that the modified Zerilli-Armstrong bcc relation resulted in an increase in the range of applicability and accuracy of the simulations [143].

In 1994 Holt et. al [144] conducted an experimental-computational investigation into the deformation behaviour of titanium during Taylor cylinder impact tests. The material behaviour of titanium in the computational aspect of the investigation was modelled using the Zerilli and Armstrong [128] constitutive relation for bcc metals. Although titanium is hcp, it is known to exhibit a bcc type of behaviour [144]. That is, similar to bcc materials, hcp materials show a strong temperature and strain rate

dependence of the flow stress. The values of the constants in the bcc equation that were used to model titanium were derived from experimental tensile data on pure titanium [145]. In those experiments [145], tests were performed at several temperatures and strain rates ranging from 1×10^{-4} to 1 s^{-1} . In order to adequately simulate titanium in the impact tests, Holt et. al [144] introduced a threshold stress for the onset of twinning which increased the calculated flow stress. It was argued that this was necessary since the bcc relation was fit to low strain rate data, where the effects of twinning were less significant [144]. They stated that slip was the preferred deformation mechanism, but at high strain rates the stress level needed to produce slip may become larger than that associated with deformation twinning.

Recognising that hcp materials exhibit deformation behaviour similar to both bcc and fcc materials, Zerilli and Armstrong [123, 130] introduced a constitutive relation that was a combination of their previous constitutive relations for bcc and fcc materials [127, 128]. The equation introduced for hcp materials was,

$$\bar{\sigma} = \sigma_a + B e^{-\beta \bar{\epsilon}} + B_0 \bar{\epsilon}^{1/2} e^{-\alpha \bar{\epsilon}} \quad (1.33)$$

where the variables are defined as above.

Zerilli and Armstrong [129, 130] applied the hcp equation to the hcp titanium alloy, Ti-6Al-4V, and the bcc alloy steel, HY-100. The deformation behaviour of Ti-6Al-4V was bcc typical, in which the Peierls stress interactions are predominant, and the HY-100 showed a significant dislocation intersection interaction, typical of fcc metals. The calculated predictions were compared to experimental data and overall the Zerilli

and Armstrong hcp relation compared very well for the titanium and steel alloys [129, 130].

1.6 Current Research

Based on the preceding review, it is evident that a limited amount of quasi-static data does exist describing the deformation behaviour of Zr-2.5Nb pressure tube material in the longitudinal direction. It is also evident that there is essentially no data describing the high strain rate deformation of pressure tube material. In order to develop constitutive relations capable of modelling the behaviour of pressure tubes during postulated accident scenarios in CANDU reactors, high strain rate data is required. The current research was undertaken to address this lack of data.

In the review of experimental techniques for high strain rate testing, it was shown that the most accepted method for determining material properties at strain rates in the range of 1×10^2 to $1 \times 10^4 \text{ s}^{-1}$ is the split Hopkinson bar technique. As a result, a direct-tension version of the split Hopkinson bar was constructed [146] and used to acquire stress-strain data over the applicable range of strains, temperatures, strain rates and irradiation fluences. Experiments were performed on specimens manufactured from both the longitudinal and transverse tube orientations, and the results were used to characterise the mechanical behaviour.

A second aspect of the current research was to use the high strain rate data obtained to characterise the constitutive behaviour of Zr-2.5Nb pressure tube material in the longitudinal direction. The review of constitutive relations showed that the

Johnson-Cook and Zerilli-Armstrong hcp relations are suitable for modelling the high strain rate deformation of pressure tube material. Therefore, constitutive parameters for these models were derived.

The Johnson-Cook and Zerilli-Armstrong hcp constitutive relations were implemented into an explicit, dynamic finite element code and used to describe the material behaviour in numerical simulations of direct-tension split Hopkinson bar experiments on Zr-2.5Nb pressure tube material. The results from the numerical simulations were compared with data obtained from the experiments to determine the predictive capability. The implementation of the constitutive relations allows results from this research to be applied to future studies involving the high strain rate deformation of Zr-2.5Nb pressure tubes and other hcp materials.

The remainder of this thesis is organised as follows. Chapter 2 provides a description of the experimental method, which includes details of the Zr-2.5Nb pressure tube material used in this investigation. The experimental results and analysis are presented in Chapter 3 and the details of the constitutive modelling are provided in Chapter 4. The numerical model and results of the simulations are outlined in Chapters 5. The results of this research are discussed in Chapter 6, and the conclusions and recommendations for future work are presented in Chapter 7.

Chapter 2

Experimental Method

The experimental component of this investigation involves the testing of as-received and irradiated Zr-2.5Nb pressure tube material at high rates of tensile strain. These tests were conducted at the AECL, Whiteshell Laboratory on a tensile split Hopkinson bar apparatus [146] that was based on the direct-tension design introduced by Staab and Gilat [118]. The direct-tension split Hopkinson bar used is capable of testing materials at strain rates ranging from approximately 500 to 3000 s⁻¹ and temperatures from room temperature to approximately 300°C [146]. A limited number of quasi-static tensile experiments were also conducted at strain rates ranging from 1 x 10⁻⁴ to 0.1 s⁻¹ at AECL, Whiteshell Laboratory using an servo-hydraulic test frame, and at high strain rates using the tensile split Hopkinson bar at Carleton University [152 - 154].

The material tested in this investigation is described in Section 2.1 and Section 2.2 provides details of the direct-tension split Hopkinson bar apparatus. Section 2.3 describes the specimen design and a description of the experimental instrumentation is provided in Section 2.4. Section 2.5 provides details on the elevated

temperature tests and the methods associated with the reduction of data are provided in Section 2.6.

2.1 Material

All of the specimens tested in this investigation were machined from actual Zr-2.5Nb pressure tubes that were removed from operating CANDU reactors. As these pressure tubes were manufactured for in-reactor use, the typical fabrication process, microstructure, crystallographic texture, deformation mechanisms, and mechanical properties, reviewed in Section 1.2, provide a detailed description of the tubes used in this investigation. Therefore, this section only describes details of the specific tubes from which the specimens were manufactured. The chemical specifications for all of the pressure tubes used in this research are included in Table 2.1 [10].

2.1.1 As-Received Zr-2.5Nb Pressure Tube Material

As-received material tested in this investigation was taken from a pressure tube removed from Pickering Nuclear Generating Station (NGS) B, Unit 8, tube M04 (P8M04) before the reactor went critical [147]. Pressure tube P8M04 was manufactured from a double-arc melted ingot #202653Q and had a total of 26.8% cold work [147, 148].

2.1.2 Irradiated Zr-2.5Nb Pressure Tube Material

The irradiated material tested in this investigation was taken from sections of pressure tubes removed from Pickering NGS A, Unit 3-tube J09 (P3J09), Bruce NGS A, Unit 2-tube N06 (B2N06), and Pickering NGS A, Unit 4-tube F17 (P4F17). Tables 2.2 to 2.4 outline the average temperatures during their in-reactor service and total fluences along the length of the pressure tubes for tube P3J09, B2N06, and P4F17, respectively [149]. Pressure tubes P3J09 and B2N06 were manufactured from ingots that were double-arc melted, and tube P4F17 was manufactured from a quad-melted ingot [149]. The ingot numbers that the pressure tubes were manufactured from, year the tubes were removed, inlet end identification, and effective full-power hours (EFPH) that the pressure tubes operated are given in Table 2.5 [147, 149]. The sections of pressure tubes used in this investigation [150, 151] were identified by the tube they were taken from and the position along the tube, as outlined in Table 2.6. The sections were taken from either the east (P-E), or west (P-W) portion of the tubes, as identified at the time of removal from the reactors. The axial position of each section, referenced from the inlet end of each tube, is included in Table 2.6 and was used to calculate the fluence levels.

2.2 Direct-Tension Split Hopkinson Bar Apparatus

As described in Section 1.4, independent of the mode of loading, the split Hopkinson bar technique involves placing a short specimen between an input and output bar. The specimen is loaded by an elastic stress wave that is generated in the input bar,

known as the incident wave. Depending on the mode of loading and specific configuration, the incident wave can be generated in a number of ways. For the direct-tension configuration of split Hopkinson bar used in this investigation [118, 146], the incident wave is generated by quickly releasing a stored load in the end of the input bar. A schematic of the direct-tension split Hopkinson bar used is shown in Figure 2.1 and a photograph of the split Hopkinson bar at AECL is provided in Figure 2.2. The split Hopkinson bar at AECL is made of 12.7-mm-diameter, 7075-T651 aluminum bar and is approximately 5.6-m-long. This allows constant strain-rate tests in excess of 600 μs to be performed. The input bar is 3.82-m-long and the output bar is 1.83-m-long, as indicated in Figure 2.1. The input load is stored by clamping the input bar at a location 2.2 m back from the sample and placing an axial load at the end. Photographs of the clamp/release and loading mechanisms are shown in Figures 2.3 and 2.4, respectively. When the clamp is released, an incident tensile wave is formed and propagates down the input bar towards the specimen. The clamp/release mechanism, Figure 2.3, consists of grips, a c-clamp, a fracture-pin and a load-bearing device. In order to place a tensile load on the input bar, the bar is first clamped by placing a load on the large c-clamp. The c-clamp squeezes the bottom of the grips and the fracture-pin holds the grips on the top, thereby clamping the input bar. The input bar is then loaded axially by placing an applied load on the end using the loading mechanism, Figure 2.4. The grips translate axially along a linear bearing and contact the load-bearing device at two points, with 6.35 mm ball bearings. The load-bearing device can rotate in the horizontal plane so that there is no bending moments placed on the input bar. Once the test is to be performed, the c-clamp is used to further load the grips to a point where the fracture-pin breaks. The grips

that clamp the bar then fly open and release the stored tensile load. The fracture pins are manufactured from 7075-T651 aluminum and are designed to break at a pre-determined load.

All of the experiments conducted on the direct-tension split Hopkinson bar at AECL were performed in the Materials Testing Laboratory (Room 1-090, Building 300). As this lab is typically not used for testing irradiated material, radiation work procedures [155] were developed that defined the operating limits and conditions for conducting high strain rate tensile tests on irradiated material outside the AECL Whiteshell Shielded Hot-Cells Facility (WL-HCF). These procedures include details in the following four main areas.

1. Specimen acquisition and laboratory preparation
2. Specimen transportation
3. Specimen testing
4. Post-test decontamination

Included in the work procedures is a description of the radiation safety measures and controls that were followed by operating staff for the expected radiological hazards.

2.3 Specimen Design

The design of the Zr-2.5Nb pressure tube specimens used in this investigation is shown schematically in Figure 2.5. The geometry of the specimen was chosen because

of the limitations associated with testing at high strain rates, as described in Section 1.4.2. Also, it was desired to have specimens of the same geometry manufactured from longitudinal and transverse tube orientations, and for as-received and irradiated specimen [146]. The specimens were manufactured by first milling a pressure tube blank into a flat 1.5-mm-thick plate and then machining the plates to the final dimensions. A TREE Journeyman model 425 milling machine was used to machine the as-received specimens and a Charmilles Technologies Eleroda model 110 spark erosion machine was used to machine the irradiated specimens. The entire manufacturing process for the irradiated specimens was performed in the WL-HCF. The specimens were fastened to the direct-tension split Hopkinson bar apparatus at AECL by way of threaded adapters attached to the specimen by press fitted pins, as shown schematically in Figure 2.5. The adapters were constructed of 7075-T651 aluminum and the pins were. The adapters were manufactured using a Colchester Electronic Triumph lathe. Photographs of the tensile specimen and the tensile specimen with adapter attached are shown in Figures 2.6 and 2.7, respectively. A photograph of the specimen attached to the direct-tension split Hopkinson bar at AECL is provided in Figure 2.8. Adapters of exactly the same dimensions and material were used for the quasi-static tests. For the tests performed on the tensile split Hopkinson bar at Carleton University, the adapters were manufactured from AISI 4340 steel, which is the same material as the Hopkinson bar. The pins attaching the specimen to the adapters were the same as that used for the AECL specimens, but the threads on the adapters were changed to match those of the Hopkinson bar at Carleton University [152 - 154].

2.4 Instrumentation

2.4.1 Strain Gauges

The incident, transmitted, and reflected strains in the direct-tension split Hopkinson bar were monitored at the locations indicated in Figure 2.1. The strains were measured with strain gauges (type WK-13-060CD-10C foil gauges) in a full Wheatstone bridge configuration. The strain gauges used were manufactured by Micro-Measurements Division of Measurements Group Inc. and had calculated gauge factors from the manufacturer of $2.10 \pm 1.0\%$. Each Wheatstone bridge consists of two gauges mounted on the Hopkinson bar and two dummy gauges which were mounted on a separate bar of 7075-T651 aluminum. This configuration compensates for any strains introduced by a change in ambient temperature. Bending moments were also monitored using a full Wheatstone bridge configuration at the end of the input bar. The bridges were powered individually with Hewlett Packard DC power supplies (model E3611A) adjusted to provide 10.0 ± 0.01 V. The signals from the bridges were recorded using a four channel Nicolet Pro-42 digital storage oscilloscope, at a sampling rate of 2×10^6 to 5×10^6 samples per second. During a test on the split Hopkinson bar 8,000 data points were recorded. The Nicolet Pro-42 digital storage oscilloscope has a temporal accuracy of 1 ns and provides 12 bits of resolution.

The recorded voltage output from the gauges, as a function of time, were converted to strain histories using the non-linear form of the Wheatstone bridge,

$$\frac{E_0}{E} = \frac{F \varepsilon_{bar} \times 10^{-3}}{2 + F \varepsilon_{bar} \times 10^{-6}} \quad (2.1)$$

where E_0 is the measured voltage across the Wheatstone bridge, E is the supply voltage, F is the gauge factor, and ε_{bar} is the strain in the bar.

2.4.2 Strain Gauge Calibration

Static calibration of the strain gauges on the split Hopkinson bar was performed a number of times during the course of the experimental work to ensure the accuracy of the strain measurements from all of the strain gauges. The calibration was performed by attaching a calibrated MTS load-cell to the end of the output bar, connecting the input and output bars, and statically loading the entire split Hopkinson bar apparatus in tension with the loading mechanism at the end of the input bar. Output from the Wheatstone bridges were recorded at various loads up to, and in excess of that experienced during an actual test and used to calculate the load on the bar. This calculated load was compared to that recorded on the MTS load cell. The output from the Wheatstone bridges was converted to strain (ε_{bar}) by Equation 2.1. The strain at each gauge location was then used to determine the load on the bar (F_{bar}) by,

$$F_{bar} = A_{bar} E_{bar} \varepsilon_{bar} \quad (2.2)$$

where A_{bar} and E_{bar} are the cross-sectional area and elastic modulus of the split Hopkinson bar, respectively. Measurements based on the strains in the split Hopkinson bar were generally within 1% of that recorded on the load-cell.

In 1988, Safford [156] questioned the validity of such static calibrations since the pulses generated in an actual test are dynamic in nature. A common method to accommodate a dynamic calibration for Hopkinson bars that use a striker to impart the incident load is to compare the expected strain for a given striker velocity [8]. Since the direct-tension version of the Hopkinson bar apparatus does not utilise a striker to load the input bar, this method could not be used. Therefore, no dynamic calibration was conducted. However, during commissioning tests of the split Hopkinson bar apparatus at AECL [146], comparisons were made with published data [3, 118]. Also, in this investigation tests were performed using the split Hopkinson bar at Carleton University. These results were compared with tests conducted on the split Hopkinson bar at AECL and showed close agreement.

2.5 Elevated Temperature Tests

The heater used to heat the specimens for tests conducted on the direct-tension split Hopkinson bar at AECL is a fully programmable Quad-Elliptical Radiant Heating Chamber, model E-4-5528-2 (manufactured by Research Inc.), as shown in Figures 2.9 and 2.10. It provides uniform circumferential heating over an effective heating zone of approximately 191 cm³ (12 cubic-inches). The outer shell is constructed of aluminum and surrounds four 1 kW quartz heating lamps with tungsten emitters. The inner portion

of the shell behind each lamp has an elliptical shape that is polished to reflect light into the heating zone. The chamber is capable of maintaining steady-state temperatures of approximately 1100°C in the heated zone. The heater is cooled by a water-cooling system contained between the inner and outer shells.

Two cylindrical shields were constructed of 7075-T651 aluminum to shield the specimen adapters and the ends of the Hopkinson bars from heating, as shown in Figures 2.9 and 2.10. When a test is to be performed, the shields are inserted into the heater through the ends, as shown in Figure 2.9. The shields reflect light away from the ends of the input and output bars, thereby minimising the heating of the adapters and the end of the bars.

For the elevated temperature tests the radiant heating chamber was used to heat the specimens and the temperature was controlled and monitored by a type-K thermocouple that was attached to the specimen. The thermocouple, however, was not attached to the center of the gauge length since this could effect the deformation behaviour. Because of temperature gradients along the specimens gauge length a relationship between the temperature at the center of the specimen, where the deformation would occur, and the location where the temperature was controlled was required. This relationship and the associated error in the measured temperature are reported in [157] and were used in this research.

For tests performed at 71 and 113°C, a thermocouple that was in contact with the center of the specimen was used to control and monitor the specimen temperature. The errors associated with this configuration were $\pm 7^{\circ}\text{C}$ and $\pm 24^{\circ}\text{C}$ for the tests performed at 71 and 113°C, respectively [157]. For the tests performed at 290°C, a thermocouple was

spot-welded to end of the gauge length and a relationship determined in [157] was used to determine the temperature at the center of the gauge length. The calculated error in the temperature at the center of the specimen for these tests was $\pm 18^{\circ}\text{C}$ [157].

2.6 Data Reduction

As described in Section 2.4.2, the recorded voltage output from the Wheatstone bridges was converted to strain histories for the gauge locations indicated in Figure 2.1. The three strain histories for gauges located at locations A, B, and C in Figure 2.1 represent the incident, reflected, and transmitted strain histories, respectively. These strain histories were applied to the Hopkinson equations and used to determine the engineering stress, strain and strain rate histories for each test specimen. These calculations were performed with a FORTRAN program (SHB_ANALYSIS) on a PC workstation. The source code for SHB_ANALYSIS is included in Appendix A.

The true-stress, true-strain, true-strain rate and specimen temperature histories were also calculated in SHB_ANALYSIS. As described in Section 1.4.2, data obtained in a tensile test performed on a split Hopkinson bar is valid between the time that deformation equilibrium is initially reached and when localised deformation (necking) occurs. Data obtained prior to deformation equilibrium cannot be used and data beyond the point where necking starts cannot be used directly to determine the true-stress and true-strain. The details of the calculations used in SHB_ANALYSIS to calculate the true properties are described in the following sections.

2.6.1 Calculation of True-Stress

Engineering stress (σ_{eng}) is calculated by,

$$\sigma_{eng} = \frac{F}{A_0} \quad (2.3)$$

where F is the applied force and A_0 is the original cross-sectional area of the gauge length. The true-stress (σ) is calculated by,

$$\sigma = \frac{F}{A} \quad (2.4)$$

where A is the true cross-sectional area. Therefore, in order to calculate the true-stress, the true cross-sectional area during the test must be known. From Equations 2.3 and 2.4 it follows that,

$$\sigma = \frac{A_0}{A} \sigma_{eng} \quad (2.5)$$

In the present investigation, σ_{eng} is calculated from the Hopkinson equations presented in Section 1.4.2, and A was determined from,

$$A = A_0(1 - RA) \quad (2.6)$$

where RA is the reduction in cross-sectional area taken to be a function of engineering strain (ϵ_{eng}). The value of A_0 was determined from pre-test measurements and RA was obtained by fitting a polynomial function to data from high strain rate tests performed on as-received and irradiated Zr-2.5Nb pressure tube material. This is a similar approach used in [153, 154]. Calculations of the reduction in cross-sectional area were based on pre- and post-test measurements of the tensile specimens, and the assumption that the volume of the gauge length remained constant until the point where necking occurred. Figures 2.11 and 2.12 show the reduction in cross-sectional area data for the as-received and irradiated Zr-2.5Nb pressure tube material as a function of engineering strain, respectively. As shown in the figures, reduction in cross-sectional area data from room temperature and 290°C tests were used to determine the polynomial functions. The data points are from experiments conducted at strain rates ranging from approximately 500 to 3000 s⁻¹. For the as-received material, the room temperature and 290°C test data were combined to determine the relationship, as shown in Figure 2.11. The relation derived for the as-received material is,

$$RA = 89.5105\epsilon_{eng} - 108.7085\epsilon_{eng}^2 + 1813.2507\epsilon_{eng}^3 \quad (2.7)$$

with $R^2 = 0.9838$. For the irradiated material, a separate relationship was derived for the room temperature and the 290°C data since the deformation behaviour, as characterised by the reduction in cross-sectional area as a function of engineering strain, appeared to be dependent on temperature, as shown in Figure 2.12. The relationship derived for the irradiated material at room temperature is,

$$RA = 67.2937\epsilon_{eng} + 670.3272\epsilon_{eng}^2 \quad (2.8)$$

with $R^2 = 0.9658$, and at 290°C it is,

$$RA = 72.7778\epsilon_{eng} + 380.429\epsilon_{eng}^2 \quad (2.9)$$

with $R^2 = 0.9956$.

Included in Figure 2.11 are data points obtained from interrupted tests performed on as-received Zr-2.5Nb pressure tube material with the tensile split Hopkinson bar at Carleton University. These tests were performed at room temperature and a strain rate of approximately 1700 s^{-1} . In these tests, the specimens were strained to a predetermined value of strain and then the tests were interrupted using a momentum trapping technique, as described in [153, 154]. Since the test specimens do not fracture when the momentum trapping technique is used, the cross-sectional area can be accurately measured for a known level of strain after the test. Therefore, these tests provide a reliable check on the accuracy and method of obtaining the relationship for the reduction in area as a function of engineering strain, as given by equation 2.7. Such a check for the irradiated material was not possible since irradiated material could not be tested at Carleton University. However, the same method of obtaining the relationship between the reduction in cross-sectional area and engineering strain was used for both as-received and irradiated material. It should be noted that equations 2.7 to 2.9 are only applicable over the range of strains used in their derivation.

2.6.2 Calculation of True-Strain

The true-strain is defined as [158],

$$\varepsilon = \ln\left(\frac{l}{l_0}\right) \quad (2.10)$$

where l is the current, or instantaneous length, and l_0 is the original length of a test specimen. For plastic flow, it can be assumed that a material is incompressible [158] and the following relationship holds,

$$\varepsilon_x + \varepsilon_y + \varepsilon_z = 0 \quad (2.11)$$

where ε_x , ε_y , and ε_z , represent the strains in the directions associated with the initial length, width and thickness of a specimen, respectively. Combining equations 2.10 and 2.11 yields,

$$\varepsilon_x + \ln\left(\frac{y}{y_0}\right) + \ln\left(\frac{z}{z_0}\right) = 0 \quad (2.12)$$

where y and z represent the instantaneous width and thickness, respectively, and y_0 and z_0 represent the original width and thickness, respectively. Since the initial and instantaneous cross-sectional areas of the specimen are calculated as,

$$A_0 = y_0 z_0 \quad (2.13)$$

and

$$A = yz \quad (2.14)$$

Equation 2.12 can be expressed as,

$$\varepsilon_x = -\ln\left(\frac{A}{A_0}\right) \quad (2.15)$$

Equation 2.15 represents the true-strain in the axial direction of a specimen, in terms of the initial and instantaneous cross-sectional areas. In the current investigation, the value of A_0 was determined from pre-test measurements and A was determined from the relation given in Equation 2.6 and described in Section 2.6.1.

2.6.3 Calculation of True-Strain Rate

In a tensile split Hopkinson test, the strain rate is approximately constant during the time that there is uniform deformation. Beyond the point where localised deformation occurs, however, local strain rates increase rapidly and can reach very high values [8, 111]. Therefore, in the current investigation the value of true-strain rate was calculated based on the instantaneous value of true-strain as,

$$\dot{\varepsilon} = \frac{\partial \varepsilon_x}{\partial t} \quad (2.16)$$

where t represents time and ε_x is calculated from equation 2.15.

2.6.4 Calculation of Specimen Temperature

As described in Section 1.3.3, testing materials at high rates of strain can lead to significant increases in temperature as a result of adiabatic heating. The increase in temperature during adiabatic deformation can be calculated by,

$$(1.7) \quad \Delta T = \frac{\beta}{\rho c_p} \int_{\varepsilon_p=0}^{\varepsilon_p=\varepsilon'} \sigma d\varepsilon_p$$

For the present investigation, the fraction of plastic work converted to heat (β) was taken to be 0.9, the density (ρ) equals 6.530 kg/m³ [13], and the specific heat (c_p in units of J/g°C) was calculated as a function of temperature from,

$$c_p = 0.2867 + 1.13714 \times 10^{-4} T \quad (2.17)$$

Equation 2.17 was derived by fitting a linear equation to published data on the specific heat of as-received Zircaloy-2 tubing [159] as a function of temperature. The temperature range for the published data [159] was 50 to 700°C.

2.6.5 Calculation of Elastic Modulus

The elastic modulus (E) is required in the calculation of the true plastic strain (ε_p), which is determined from [158],

$$\varepsilon_p = \varepsilon_{total} - \varepsilon_e = \varepsilon_{total} - \frac{\sigma_{eng}}{E} \quad (2.18)$$

where ε_{total} is the total strain ($\equiv \varepsilon_x$), and ε_e is the elastic component of strain determined by,

$$\varepsilon_e = \frac{\sigma}{E} \quad (2.19)$$

In the present investigation, E was taken to be a function of temperature and pressure tube orientation. For Zr-2.5Nb pressure tube material loaded in the longitudinal direction, the associated elastic modulus (E_{long} in units of GPa) was calculated by [160],

$$E_{long} = 965633 - 1.40282 \times 10^{-5} T - 2.91282 \times 10^{-4} T^2 + 3.87923 \times 10^{-7} T^3 \quad (2.20)$$

Chapter 3

Experimental Results and Analysis

This chapter presents the results of experiments performed on as-received and irradiated Zr-2.5Nb pressure tube material using the direct-tension split Hopkinson bar at AECL. These experiments were performed on specimens manufactured from the longitudinal and transverse directions at temperatures ranging from 20°C, denoted as room temperature (RT), to approximately 300°C and strain rates from approximately 500 to 2500 s⁻¹. The fluence level of the irradiated material ranged from approximately 1 x 10²⁵ to 8 x 10²⁵ n/m². The presentation of the results in this chapter concentrates on engineering stress-strain data, since it directly represents the experimentally measured values and allows for a direct comparison of data obtained in a previous investigation [81]. The manipulated data in the form of true-stress and true-strain is presented in Chapter 4.

As part of the current research, a limited number of tensile tests were also performed on as-received material at quasi-static strain rates at AECL, and at high strain rates at Carleton University with the split Hopkinson bar apparatus. Results from the

quasi-static tests (Section 3.1) provide a basis for comparison with the high strain rate data and extend the current data base on the tensile properties of Zr-2.5Nb pressure tube material. The results from the Hopkinson bar tests at Carleton University were used to validate the calculations of true-stress and true-strain, as outlined in Sections 2.6.1 and 2.6.2. The high strain rate experimental results from tests performed at AECL and Carleton University are presented in Sections 3.2 and 3.3, respectively. Observations of the deformation behaviour of tested specimens are presented in Section 3.4.

3.1 Quasi-Static Results

Quasi-static tests were performed on specimens manufactured from as-received pressure tube material using a servo-hydraulic testing frame at AECL, Whiteshell Laboratory as per ASTM-Designations E 8 and E 21 [161]. Tests were performed at strain rates of 1×10^{-4} , 1×10^{-3} , 1×10^{-2} and $1 \times 10^{-1} \text{ s}^{-1}$, and at room temperature, 250°C, and 300°C for both longitudinal and transverse specimens. The servo-hydraulic testing frame was equipped with a laser extensometer system (Zygo Laser Telemetric System, model 121) to measure the extension on both sides of the specimen. The voltage output from the extensometers were converted to elongation (mm) as a function of time and averaged to yield the specimen elongation versus time, which was used to calculate the engineering strain. The load history was obtained by recording the voltage output from the load-cell as the experiments were conducted and converting to load as a function of time, which was used to calculate the engineering stress.

Tables 3.1 and 3.2 outline all of the quasi-static tests performed on the longitudinal and transverse specimens, respectively. Appendix B contains the engineering stress-strain responses for the longitudinal specimens and Appendix C contains the responses for the transverse specimens. The laser extensometer system was used to measure the elongation of the specimens for all of the quasi-static tests, except those performed at 300°C and strain rates of 1×10^{-2} and $1 \times 10^{-3} \text{ s}^{-1}$. For these experiments, the elongation as a function of time was estimated by the cross-head displacement of the apparatus.

3.1.1 Longitudinal Results

Figure 3.1 shows typical engineering stress-strain responses from longitudinal specimens tested at room temperature and 300°C, for strain rates ranging from 1×10^{-4} and $1 \times 10^{-1} \text{ s}^{-1}$. From Figure 3.1 it is seen that the strain hardening behaviour of the longitudinal specimens over this range of strain rate and temperature is quite similar. The average σ_{UTS} increased 52 MPa with an increase in the rate of strain from 1×10^{-4} to $1 \times 10^{-1} \text{ s}^{-1}$ at room temperature, and 41 MPa at 300 °C. The overall effect of increasing the temperature was to decrease the strength. For a temperature increase from room temperature to 300°C, the average σ_{UTS} decreased by 191 MPa for specimen tested at $1 \times 10^{-4} \text{ s}^{-1}$, and 202 MPa for specimens tested at $1 \times 10^{-1} \text{ s}^{-1}$, which represents a decrease of 26% for both strain rates. Also, there appears to be a slight increase in ductility for increasing strain rate and temperature, as indicated by the failure strains outlined in Table 3.1. These results are comparable to those obtained by Ibrahim [81]. In that

investigation, quasi-static tensile tests showed that the σ_{UTS} of the as-received Zr-2.5Nb pressure tube material was 549 ± 30 MPa at 300°C and the elongation at fracture was $17 \pm 2.6\%$. At room temperature, the σ_{UTS} increased by approximately 250 MPa [81].

3.1.2 Transverse Results

Figure 3.2 shows typical engineering stress-strain responses for transverse specimens tested at room temperature and 300°C for strain rates ranging from 1×10^{-4} and $1 \times 10^{-1} \text{ s}^{-1}$. Comparing Figures 3.1 and 3.2 it is seen that the overall stress-strain behaviour for the transverse specimens is quite different than that of the longitudinal specimens. As observed in previous investigations on Zr-2.5Nb pressure tube material [73, 74, 81-83], the transverse specimens did not strain harden gradually, but rather increased in strength rapidly reaching the point of σ_{UTS} very close to the 0.2% yield point. In the room temperature tests, the strength remained almost constant from this point to approximately 10 % strain and then dropped off, or strain softened, to the point of fracture. In the 300°C tests, the specimens strain softened steadily after σ_{UTS} to the point of fracture.

The difference in the stress-strain behaviour of the longitudinal and transverse specimen is due to the anisotropic nature of the pressure tube material, as described in Section 1.2.5. The crystallographic texture of the pressure tube material causes the longitudinal specimen to deform primarily by slip, whereas the transverse specimen deform primarily by twinning. As a result, more work hardening is observed in the longitudinal specimens.

The transverse specimens exhibited an increase in strength with an increase in strain rate over the quasi-static regime, as seen in longitudinal specimens. The average increase in σ_{UTS} for the room temperature transverse specimens was 105 MPa for an increase in strain rate from 1×10^{-4} to $1 \times 10^{-1} \text{ s}^{-1}$. An increase of 40 MPa was observed in the 300°C tests for the same increase in strain rate. As observed with the longitudinal specimens, increasing the temperature from room temperature to 300°C caused a decrease in the strength. At a strain rate of 1×10^{-4} the strength decreased 194 MPa (24%) and at a strain rate of $1 \times 10^{-1} \text{ s}^{-1}$, the strength decreased 259 MPa (29%). Also, the average σ_{UTS} values for the transverse specimens measured in this investigation are comparable to those determined in [81], where the average σ_{UTS} was 604 ± 30 MPa at 300°C and increased by approximately 250 MPa for tests conducted at room temperature. In contrast to the longitudinal results, there appears to be little effect of strain rate or temperature on ductility for the transverse specimens, as seen in the failure strains outlined in Table 3.2.

Comparing the quasi-static test results from the longitudinal and transverse specimens, it appears that the transverse specimens are stronger than the longitudinal specimens for the range of temperatures considered. From the results outlined in Tables 3.1 and 3.2, it is seen that pressure tube material is approximately 12 to 18% stronger in the transverse direction for the range of strain rates and temperatures considered. This observation is consistent with the quasi-static results from [81] where it was reported that as-received pressure tube material was an average of 10% stronger in the transverse direction for temperatures between room temperature and approximately 300°C. It should also be noted that the transverse specimens exhibit strong strain

softening behaviour, relative to the longitudinal specimen. This behaviour suggests that the pressure tube material has a lower mechanical stability in the transverse direction, as reflected by the lower ductility.

3.2 AECL High Strain Rate Results

One of the advantages of the direct-tension version of the tensile split Hopkinson bar apparatus [118] is that a relatively constant strain rate is obtained during most of the duration of a test. For all of the high strain rate tests in this investigation, the stated value of strain rate was calculated by averaging the raw strain rate data for measured strains from 2% to failure. Figure 3.3 shows typical strain rate as a function engineering strain data from two high strain rate tests. One test (E50504A) was conducted at an average strain rate of 440 s^{-1} and the other (E50508C) at 2650 s^{-1} . In both cases, it can be seen that the strain rate rises sharply and then levels off to a near constant value. The accumulated strain at the point where a constant strain rate is obtained depends on the rate of strain. For the lower strain rate test (E50504A), a constant value of strain rate is obtained at an accumulated strain level below approximately 1%. For the higher strain rate test (E50508C) a constant strain rate is reached later, at approximately 4 to 5% engineering strain. As the majority of the deformation takes place beyond these strain levels, it can be assumed that the high strain rate tests were indeed performed at a near constant rate of strain. It should be noted that high strain rate data is generally not suitable for the determination of initial yield stress as a result of this initial transient behaviour.

3.2.1 As-Received Material

All of the high strain rate tests performed on the as-received longitudinal and transverse specimen using the direct-tension split Hopkinson bar at AECL are summarised in Tables 3.3 and 3.4, respectively. Tests were performed at room temperature, 71, 113 and 290°C. Appendix D contains the engineering stress-strain responses for the tests on the longitudinal specimens and the engineering stress-strain responses for the transverse specimens are included in Appendix E.

Longitudinal High Strain Rate Results

Based on all of the high strain rate tests on as-received longitudinal specimen, the overall stress-strain behaviour is similar to that observed at quasi-static strain rates, except at the highest strain rates. That is, the longitudinal specimens exhibit steady work hardening to the point of σ_{UTS} and then gradual strain softening to the point of failure. However, as the strain rate increases over the high strain rate regime, there appears to be a transition in the initial work hardening behaviour. This trend is depicted in Figure 3.4 and shows that the stress increases at a near constant rate to the point of σ_{UTS} for the highest strain rate tests (E50508C and E60312B), which is at a noticeably lower strain level. This behaviour suggests that the initial work hardening rate is strain rate sensitive and increases as the strain rate is increased. However, this observation is only qualitative due to the transient nature of a high strain rate test, as noted above and discussed in detail in Section 4.1.

In recent compression tests on pure zirconium [59, 60] it was observed that the rate of strain hardening increased with increasing strain rate and decreasing temperature. The observed rate of strain hardening in [60] was much greater when the specimens were loaded in a direction perpendicular to the hcp *c*-axis. This orientation is equivalent to the longitudinal specimen in the present investigation and is aligned preferentially for slip-dominated deformation.

There is a decrease in strength with increasing temperature at high strain rates, as was seen in the quasi-static tests on longitudinal specimens. The overall effect of strain rate and temperature on the strength of as-received Zr-2.5Nb pressure tube material in the longitudinal direction is shown in Figure 3.5. From Figure 3.5 it is seen that there is an increase in strength with increasing strain rate for temperatures ranging from room temperature to 300°C. Also, it appears that the effect of strain rate is more pronounced in the high strain rate regime, which is in agreement with the bi-linear rate sensitivity observed for α -titanium [7].

The failure strains for the longitudinal specimens are outlined in Tables 3.1 and 3.3, and plotted as a function of logarithm of strain rate in Figure 3.6. Based on these results, there appears to be a decrease in ductility with increasing strain rate, from quasi-static to high strain rates over the temperature range considered. At room temperature, the failure strains of specimens tested at quasi-static rates ranged from approximately 28 to 33%, while at high rates the failure strains ranged from 20 to 25%. Averaging all of the data for the quasi-static and high strain rate tests yields failure strains of 30% and 23%, respectively. Therefore, the reduction in failure strain at room temperature, due to increasing the strain rate from quasi-static to high strain rates, is

approximately 23%. At elevated temperatures, a similar trend is observed and the average reduction in failure strain due to this increase in strain rate is approximately 21%.

Transverse High Strain Rate Results

Figure 3.7 shows typical high rate engineering stress-strain responses obtained for transverse specimens manufactured from as-received material and tested at room temperature and 290°C. As seen for the longitudinal specimens, the overall strain hardening behaviour for the transverse specimen tested at high strain rates is very similar to that observed for the transverse specimens tested at quasi-static rates. This implies that the primary deformation mechanism in the transverse high strain rate specimens is also twinning. However, for specimens tested at the highest strain rates (E50515B and E60219C) the stress increases steeply to the point of σ_{UTS} and then drops abruptly before strain softening to failure. Upon examination of the stress strain responses from all of the high strain rate tests on transverse specimens (Appendix E), it appears that this behaviour occurs proportionally as the strain rate increases. It is speculated that there is a higher propensity of deformation twinning due to increasing strain rate in the transverse specimens. This behaviour has been observed in pure zirconium specimens tested in compression [59, 60]. In these investigations [59, 60], an increase in the amount of deformation twinning was caused by an increase in the rate of strain, or a decrease in temperature. Similar results were also observed for pure titanium [58] and the titanium alloy, Ti-6Al-4V [56, 57].

Figure 3.8 shows the effect of strain rate and temperature on σ_{UTS} for the as-received material in the transverse direction. As observed for the longitudinal specimens, there is an increase in the strength due to an increase in the strain rate over the range of temperatures considered. The effect of increasing the temperature from room temperature to 300°C was to decrease the strength. Compared to the longitudinal data shown in Figure 3.5, there appears to be a more pronounced bi-linear dependence of strength on strain rate.

The failure strains of transverse specimens at quasi-static and high strain rates are summarised in Tables 3.2 and 3.4 respectively, and are shown as a function of logarithm of strain rate in Figure 3.9. At room temperature, the failure strains of specimens tested at quasi-static rates ranged from 23 to 26%, while at high rates the failure strains were lower and ranged from 15 to 20%. Averaging this data for the quasi-static and high strain rate tests yields failure strains of 25% and 17%, respectively. Therefore, the reduction in failure strain at room temperature due to increasing the strain rate is approximately 31%. At elevated temperatures, a similar trend is observed and the average reduction in failure strain was approximately 20%.

Comparing the trends in the failure strains of the transverse and longitudinal data (Figures 3.6 and 3.9), it appears that the effect of strain rate and temperature on ductility is approximately the same for both the longitudinal and transverse specimens. That is, a decrease in ductility for an increase in strain rate and decreasing temperature. It should also be noted that under all test conditions investigated, the as-received Zr-2.5Nb pressure tube material fails at lower levels of strain in the transverse direction, as compared to the longitudinal direction. Results from the high strain rate experiments

show that the strength in the transverse direction is higher than in the longitudinal direction, demonstrating that as-received Zr-2.5Nb pressure tube material remains anisotropic at high strain rates.

3.2.2 Irradiated Material

High strain rate tests were also performed on longitudinal and transverse specimens manufactured from irradiated pressure tube material, as outlined in Chapter 2. Tests were performed at temperatures ranging from room temperature to 290°C and strain rates ranging from approximately 400 to 2700 s⁻¹. The tests performed on the longitudinal specimens are summarised in Tables 3.5 and 3.6 for the room temperature and elevated temperature tests, respectively. The tests performed on the irradiated transverse specimens are summarised in Tables 3.7 and 3.8. Appendix F contains the engineering stress-strain responses for the tests on the irradiated longitudinal specimens and the responses for the transverse specimens are included in Appendix G.

Longitudinal High Strain Rate Results

Typical high strain rate engineering stress-strain responses for irradiated longitudinal specimens tested at room temperature and 290°C are shown Figure 3.10. Two of the responses shown are from tests conducted at strain rates of approximately 500 s⁻¹ and the other two are from tests conducted at approximately 2000 s⁻¹. From Figure 3.10 it is seen that the overall work hardening behaviour, as a function of strain

rate, is similarly independent of temperature. Upon examination of the tests performed on irradiated longitudinal specimens (Appendix F) it is seen that the strength increases for increasing strain rate. As seen in the as-received longitudinal specimens, there exists a transition in the work hardening behaviour and the rate of work hardening increases with increasing strain rate. Also, the strain at which the σ_{UTS} is reached, as well as the failure strain, decreases for an increase in strain rate, if the temperature is held constant. The overall effect of temperature is to decrease the strength, as expected.

Upon examination of the stress-strain responses for the as-received (Appendix D) and irradiated (Appendix F) material, it is seen that there is a large variability in the irradiated data relative to the as-received data. It is suspected that most of the variability is a result of the difficulties associated with machining the irradiated specimens and attaching them to the adapters remotely in the WL-HCF. However, some variability is expected as a result of testing material from different tubes and from different locations along the length of a single tube [81].

Although an extreme amount of care was taken in manufacturing the irradiated specimens, the use of manipulators does not permit the same level of precision handling compared to a normal workshop environment. Therefore, it is reasonable to believe that small misalignment of the specimen - adapter configuration may have occurred. Such misalignments would result in inconsistent stress-strain measurements. Also, as described in Section 2.3, the irradiated specimens were manufactured with a spark erosion machine. During machining of the irradiated specimens, it was noted that the finished surface was not smooth. The resulting geometric imperfections may have affected the failure behaviour.

Figures 3.11 and 3.12 show the effect of irradiation on longitudinal high strain rate specimens tested at room temperature and 290°C, respectively. From these figures it is seen that the main effect of irradiation is to increase strength and decrease ductility. Similar observations were made for irradiated tensile specimens tested at quasi-static strain rates [81]. Comparing the as-received and irradiated stress-strain curves, it is seen that there is an increase in the level of strain at which the σ_{UTS} is reached in the irradiated material.

The effect of strain rate and irradiation on the strength of longitudinal specimens at room temperature is shown in Figure 3.13. Comparing the data from the as-received and irradiated specimens, it appears that the overall effect of irradiation is to increase the strength for strain rates ranging from 1×10^{-4} to $3 \times 10^3 \text{ s}^{-1}$. The σ_{UTS} is increased approximately 200 MPa over this strain rate regime. The irradiated data is plotted using different symbols indicating different fluence trends to examine if there was an overall effect of fluence level. Based on this data it appears that for the range of fluences considered, there was no effect on the room temperature, longitudinal strength of Zr-2.5Nb pressure tube material. In a previous investigation on irradiated Zr-2.5Nb pressure tube material at quasi-static strain rates [81], there was a large initial increase in strength due to a fluence of approximately $0.7 \times 10^{25} \text{ n/m}^2$, after which there was no further significant increase as the fluence increased to approximately $8 \times 10^{25} \text{ n/m}^2$. This behaviour has been attributed to irradiation induced damage saturating at fluence levels of approximately $1 \times 10^{25} \text{ n/m}^2$ in zirconium alloys [13, 66, 67]. Included in Figure 3.13 are trend lines indicating the strain rate sensitivity. Although there is only one quasi-static data point for the irradiated material [81], a bi-linear dependence of strength on

log-strain rate appears to exist for the irradiated pressure tube material at room temperature.

Figure 3.14 shows the effect of strain rate and irradiation on the strength of longitudinal specimens at 290 to 300°C. The effect of irradiation is to increase the strength over the range of fluences and strain rates considered. As seen for the corresponding room temperature tests, it appears that there is no systematic effect of fluence level on the longitudinal strength at elevated temperatures, within the range of fluences considered. A bi-linear dependence of strength on log strain rate appears to exist at elevated temperatures. The overall effect of temperature on the σ_{UTS} of irradiated longitudinal specimens can be seen in Figure 3.15. There is a decrease in strength for increasing temperature and the strain rate sensitivity appears to be higher at elevated temperatures. Based on the limited quasi-static data, it further appears that the decrease in strength is approximately the same for both strain rate conditions, quasi-static and high rate.

Comparisons of the failure strains for the as-received and irradiated longitudinal specimens tested at high strain rates are shown in Figures 3.16 and 3.17 for room temperature and 290°C, respectively. Included in the figures are linear fits to the as-received and irradiated data. Since there is a limited amount of data, the linear fits merely provide a method of qualitatively examining the effect of strain rate on ductility. The failure strains for the irradiated specimens are plotted as a function of fluence using different symbols to represent different fluence levels. Based on this data, it appears that irradiation fluence does not affect the failure strain of the longitudinal specimens. This behaviour is consistent with previous investigations [13, 73, 81], which observed that the

mechanical properties of Zr-2.5Nb pressure tube material do not change significantly once the material has been irradiated to fluences of approximately of $1 \times 10^{25} \text{ n/m}^2$. There does appear, however, to be a moderate decrease in the average failure strain due to irradiation at both room temperature and 290°C. The average magnitude of the decrease in failure strain with irradiation is approximately 7% and 10% strain at room temperature and 290°C, respectively. This suggests that there is a decrease in the ductility of the longitudinal specimens due to irradiation, over the high strain rate regime considered. It also appears that there may be a slight decrease in failure strain, or ductility, for increases in the strain rate from 500 to 2500 s^{-1} .

Transverse High Strain Rate Results

Typical high strain rate engineering stress-strain curves for irradiated transverse specimens tested at room temperature and 290°C are shown Figure 3.18. As seen for the longitudinal specimens, the strength and rate of work hardening increases for increasing strain rate. Also, the strain at which σ_{UTS} and failure occur, for a given temperature, decreases for increasing strain rate and the effect of temperature is to decrease the strength. Unlike the longitudinal specimens (Figure 3.10), the strengthening due to increasing strain rate in the transverse specimens appears to be greater at room temperature.

Figures 3.19 and 3.20 show the effect of irradiation on transverse specimens tested at high strain rates. Again, the main effect of irradiation is to increase strength and decrease the ductility for the strain rate and temperature ranges considered. Similar to

the longitudinal specimens, there is an increase in the level of strain at which the σ_{UTS} is reached for the irradiated transverse specimens. Unlike the longitudinal specimens, the work hardening rate of the transverse specimens, to the point of σ_{UTS} , decreases due to irradiation.

Figure 3.21 and 3.22 show the effect of strain rate and irradiation on the strength of transverse specimens at room temperature and 290°C, respectively. As seen for the irradiated longitudinal specimens (Figures 3.13 and 3.14), the overall effect of irradiation is to increase the transverse strength for the range of fluences, strain rates, and temperatures considered. Also, a bi-linear dependence of strength on logarithm of strain rate appears to exist for the irradiated transverse specimen at both temperatures. The average increases in the transverse σ_{UTS} due to irradiation is approximately 200 and 150 MPa at room temperature and 290°C, respectively. The effect of temperature on the strength of irradiated transverse specimens is shown as a function of strain rate in Figure 3.23. As seen for the high strain rate longitudinal data, there is a large amount of scatter in the data, but there is a decrease in strength with an increase in temperature. It appears that the decrease in the transverse strength at high strain rates is approximately the same as that measured at quasi-static rates [81].

Comparisons of the failure strains for the as-received and irradiated transverse specimens tested at high strain rates are shown in Figures 3.24 and 3.25 for room temperature and 290°C, respectively. As seen for the longitudinal specimens, there appears to be a moderate decrease in the failure strain, or ductility, of transverse specimens due to irradiation, but no effect of fluence level. The decrease in failure strain due to irradiation is approximately 3% strain at both room temperature and 290°C, which

is less than the observed change for the longitudinal specimens. Also, there may be a slight decrease in transverse failure strain for increasing strain rate, although considerable scatter exists in the data. Comparing Figures 3.24 and 3.25, it appears that there is not a significant change in the ductility of as-received and irradiated pressure tube material for the temperature range considered.

Figure 3.26 provides a comparison of the effect of strain rate on strength for irradiated longitudinal and transverse specimens at room temperature. The irradiated Zr-2.5Nb pressure tube material is stronger in the transverse direction at high strain rates, which is consistent with quasi-static data [81]. At a strain rate of $1 \times 10^{-4} \text{ s}^{-1}$ the pressure tube material is approximately 100 MPa stronger in the transverse direction. Over the high strain rate regime considered in this investigation, the transverse specimens were between 150 and 250 MPa stronger than the longitudinal specimens. Although there is considerable scatter in the data, it appears that the difference in strength between the longitudinal and transverse specimen increases for increasing strain rate. This observed anisotropy is shown to also exist at elevated temperatures, as shown in Figure 3.27. The difference in the strength between the longitudinal and transverse specimens, however, is noticeably less at elevated temperatures. As seen at room temperature, it appears that the difference in strength, of the longitudinal and transverse specimens increases for increasing strain rate.

Comparing the failure strains of the longitudinal (Figures 3.16 and 3.17) and transverse (Figures 3.24 and 3.25) specimens at high strain rates, it is seen that the longitudinal specimens are more ductile for both the as-received and irradiated material. The same observation was made by [81] for pressure tube material tested at quasi-static

strain rates. It is interesting to note, however, that the difference between the failure strains of the longitudinal and transverse specimens is significantly less for the irradiated material.

3.3 Carleton University High Strain Rate Results

A limited number of high strain rate tests were conducted on as-received material at room temperature using the tensile split Hopkinson bar at Carleton University. The objective of these tests was to obtain data from interrupted tests for verifying the reduction in area calculations outlined in Section 2.6.1. Figure 3.28 shows the engineering stress-strain curves from three interrupted tests performed on longitudinal specimens. The tests were performed such that approximately 12, 13 and 19% axial strain was accumulated when the tests were interrupted. The strain rate of these tests ranged from approximately 1650 to 1800 s⁻¹. Included in Figure 3.28, is the engineering stress-strain response from a test conducted at a similar strain rate on a longitudinal specimen using the direct-tension split Hopkinson bar at AECL. It is seen that the data obtained from the split Hopkinson bar at Carleton is relatively noisy compared to the AECL data. As discussed in Section 1.4.3, this is inherent with the method used to impart the incident pulse into the input bar. Numerical techniques can be employed to smooth the data [153], but were not used in this investigation. Reduction in cross-sectional area data from the interrupted tests performed on the longitudinal specimen was previously plotted in Figure 2.11.

Figure 3.29 shows the engineering stress-strain response from an interrupted test performed at room temperature on a transverse specimen at Carleton University. The test was performed at an average strain rate of 1850 s^{-1} and was interrupted at approximately 14% strain. Included in Figure 3.29 is the engineering stress-strain response from a test conducted at a similar strain rate on a transverse specimen using the split Hopkinson bar at AECL. The data in Figure 3.29 suffers the same oscillations as observed in the longitudinal data. Comparing Figures 3.28 and 3.29 it is seen that the characteristic stress-strain response, as a function of specimen orientation, is also apparent from the Carleton data. That is, the longitudinal specimen exhibit initial hardening and then strain-soften gradually past the point of σ_{UTS} whereas the transverse specimens soften rapidly.

3.4 Deformation Behaviour

Optical microscopy was performed on a number of specimens and used to determine the reduction in cross-sectional area for the calculation of true-stress and true-strain, as described in Sections 2.6.1 and 2.6.2, respectively. The light microscope used was a WILD stereo microscope (model M400) with a SONY colour CCD-Iris/rgb video camera (model DXC-1151A). Prior to examination, the specimens were cleaned in an ultra-sonic alcohol bath to remove any debris. Figures 3.30 to 3.33 contain photographs of the longitudinal-radial and longitudinal-transverse planes for longitudinal as-received specimens tested to failure at various temperatures and strain rates.

Comparing the longitudinal-radial plane view (a) to the longitudinal-radial plane view (b) in Figures 3.30 to 3.33 shows that there is more deformation in the radial direction, than the transverse direction independent of temperature and strain rate. This behaviour is consistent with that expected due to the crystallographic structure. Since the pressure tube is highly textured with the majority of the α -zirconium grains having their *c*-axis oriented in the transverse direction, deformation perpendicular to the loading direction (radial direction) will be accommodated primarily by prismatic slip. Examining the photograph of a transverse specimen, as shown in Figure 3.34 further supports this. Since both the radial and longitudinal orientations will readily deform by prismatic slip, significant deformation is observed in both directions.

The effect of temperature on deformation can be seen by comparing the figures from longitudinal room temperature tests (RT) to 290°C tests, for quasi-static (Figures 3.30 and 3.31) and high strain rates (Figures 3.32 and 3.33). From these comparisons, there is more deformation in both the transverse and radial directions at the elevated temperatures for the quasi-static and high strain rate specimens. This observation implies that an increase in the temperature increases the ductility of the longitudinal specimens. As a result, there is a larger reduction in area (RA) for the longitudinal specimens at 290°C, as previously shown in Figure 2.11.

Comparing the photographs of the quasi-static tests (Figures 3.30 and 3.31) to the high strain rate tests (Figures 3.32 and 3.33) it appears that increasing the strain rate decreases the amount of deformation at failure, independent of temperature. That is, there appears to be more deformation prior to failure in both the transverse and radial directions for the quasi-static specimens as compared to the high strain rate specimens,

for a given temperature. This observation is also consistent with the measured strains at failure for the longitudinal specimens. In addition, Figures 3.24 and 3.25 show a slight decrease in failure strain due to increasing strain rate for the room temperature and 290°C specimen, respectively.

Figure 3.35 shows the deformation occurring in a longitudinal as-received specimen tested at room temperature with an average strain rate of approximately 1700 s^{-1} . This test was conducted on the tensile split Hopkinson bar at Carleton University and was interrupted when the total accumulated strain had reached approximately 19%. Since this test was interrupted prior to failure, a physical measurement of the pre- and post-test cross-sectional area could be performed. These results could then be used to accurately determine the reduction in area at known levels of strain. Such measurements were performed using three longitudinal test specimens and used to verify the reduction in cross-sectional area as a function of engineering strain calculations presented in Section 2.6.1.

The effect of irradiation on the deformation of longitudinal specimens can be examined by comparing the photographs of high strain rate irradiated specimens in Figures 3.36 and 3.37 with the as-received specimens in Figures 3.32 and 3.33. It is evident in the as-received specimens that there is more deformation in the radial direction than in the transverse direction for the irradiated specimens. However, it appears that the amount of deformation in both directions is reduced as a result of irradiation, independent of temperature. In a similar manner to the as-received specimens, an increase in temperature for the irradiated specimens caused an increase in the amount of deformation

in the radial direction prior to failure, while little difference was observed in the transverse direction.

Chapter 4

Constitutive Modelling

One of the principal objectives of this research is to characterise the constitutive behaviour of Zr-2.5Nb pressure tube material using a constitutive relation that is compatible with existing finite element codes currently being used for modelling pressure tube rupture scenarios. Based on the review of constitutive models presented in Section 1.5, the Zerilli-Armstrong relation for hcp materials (equation 1.33) is an appropriate choice. In its current form, this relation is based on the dislocation mechanics associated with the physics of slip deformation. Since slip is the primary mode of deformation of pressure tube material in the longitudinal direction the Zerilli-Armstrong hcp relation was used to model the longitudinal deformation. No attempt was made in the current research, however, to model deformation in the transverse direction since the Zerilli-Armstrong hcp relation is not applicable for deformation twinning. Also, the empirically-based Johnson-Cook constitutive relation (equation 1.22) was used to model the deformation of pressure tube material in the longitudinal direction to provide a comparison with results from the Zerilli-Armstrong

hcp relation. It is believed that the constitutive relations determined in this investigation will produce accurate results when applied to pressure tube rupture scenarios since recent investigations demonstrated that one of the main parameters affecting the possibility of guillotine failure was the tensile deformation of the pressure tube in the longitudinal direction [11, 12].

To determine the constitutive parameters for the as-received material, two different procedures were used. The first procedure was based on the assumptions that the strain rate and temperature remained constant throughout the test, and was denoted the isothermal - constant strain rate fit. For the second procedure, the effects of adiabatic heating and increasing strain rate were considered and the resulting fit was denoted as the adiabatic - true-strain rate fit. For the isothermal - constant strain rate fit the techniques presented in [125, 126] for the Johnson-Cook relation and in [130, 165] for the Zerilli-Armstrong hcp relation were used to determine the constitutive parameters. For the adiabatic - true-strain rate fit a non-linear least squares regression routine was used to optimise the constitutive parameters. Simultaneous non-linear regression was used to determine the parameters to within three significant figures. A commercial software package (Microcal-Origin), that utilises the Levenberg-Marquardt algorithm, was used to perform the regressions [164].

The data used for the constitutive modelling in this research is presented in Section 4.1, and the details of the methods used to determine the parameters for the as-received and irradiated material are outlined in Sections 4.2 and 4.3, respectively.

4.1 Constitutive Data

The determination as to which constitutive relation to use for a given material system is generally a semi-empirical process involving an understanding of the physical mechanisms involved in plastic deformation and the application of experimental data [162]. The experimental or constitutive data is usually obtained from tests where the strain rate and specimen temperature were assumed to be constant. These assumptions are usually incorrect since localised deformation (necking) can lead to significant increases in local rates of strain and adiabatic heating occurs as a result of plastic work. The effect of adiabatic heating can be minimised by obtaining the required data from quasi-static or incremental-loading [142] tests which are conducted under isothermal conditions. These approaches, however, do not provide the required high strain rate data [7, 162].

To demonstrate how necking and adiabatic heating influences the stress, strain rate and temperature during a high strain rate test, the true-stress, true-strain rate and specimen temperature are plotted as a function of plastic strain in Figures 4.1, 4.2 and 4.3, respectively. The data shown is from an as-received Zr-2.5Nb pressure tube specimen tested at room temperature (RT) at a nominal strain rate of 1060 s^{-1} (experiment E50327A). The equations used to calculate the true-stress, plastic strain, true-strain rate and specimen temperature are outlined in detail in Section 2.6, and are given by,

$$(2.5) \quad \sigma = \frac{A_0}{A} \sigma_{eng}$$

$$(2.18) \quad \varepsilon_p = \varepsilon_{total} - \varepsilon_e = \varepsilon_{total} - \frac{\sigma_{eng}}{E}$$

$$(2.16) \quad \dot{\varepsilon} = \frac{\partial \varepsilon_x}{\partial t}$$

where, $\varepsilon_x \equiv \varepsilon_p$ and

$$(1.7) \quad \Delta T = \frac{\beta}{\rho c_p} \int_{\varepsilon_p=0}^{\varepsilon_p=\varepsilon'} \sigma d\varepsilon_p$$

From Figure 4.1 it is seen that initially the true-stress increases rapidly and the work hardening rate is quite high. However, at approximately 5% plastic strain the work hardening rate decreases significantly and remains basically constant until failure (~ 40% plastic strain). The change in the strain rate as a function of plastic strain is shown in Figure 4.2. The nominal strain rate is that calculated if the reduction in cross-sectional area due to localisation of the deformation is not considered. Comparing the true-strain rate to the nominal strain rate it is seen that there is a considerable difference. At 40% plastic strain, the true-strain rate has increased to approximately five times that of the nominal strain rate. Similarly, the effect of adiabatic heating on the specimen temperature is shown in Figure 4.3. The actual specimen temperature, as calculated by

equation 1.7, increases by approximately 175 K at 40% plastic strain. These increases demonstrate the importance of including the effects of localised deformation and adiabatic heating when determining the constitutive parameters for high strain rate data.

The true-stress, true-strain rate and temperature were calculated at plastic strain increments of 0.002 and used for the constitutive modelling of both the as-received and irradiated material. This constitutive data was calculated for plastic strain values ranging from an initial plastic strain (ε_p^i) to a final plastic strain (ε_p^f). For the high strain rate data, the value of ε_p^i was taken to be a function of strain rate since the specimen requires a finite amount of time to reach dynamic equilibrium. The time to establish dynamic equilibrium is denoted as the rise-time and is a characteristic of the particular Hopkinson apparatus being used. The rise-time of the incident pulse for the direct-tension split Hopkinson bar used in this investigation was measured to be approximately 35 μ s. Then, as described in Section 1.4.2, π reverberations of the incident pulse within the specimen are required to reach dynamic equilibrium at which point a state of uniform deformation exists [117]. For the specimens used in this investigation, approximately 10 μ s is required, assuming an elastic wave speed of 5100 m/s and a specimen length (between pins) of 15 mm. Therefore, once the leading edge of the incident pulse reaches the specimen, approximately 45 μ s is required to reach a state of uniform deformation. Since experiments were performed at different strain rates, different amounts of strain were accumulated during this time, as shown in Figure 3.3. A functional value of ε_p^i was determined considering results from all of the experiments performed on as-received, longitudinal high strain rate specimen as,

$$\varepsilon_p^i = 0.0017 \dot{\varepsilon} + 0.6392 \text{ (\%)} \quad (4.1)$$

Based on equation 4.1, values of ε_p^i for the high strain rate data in this investigation ranged from approximately 2 to 5%. For the quasi-static data, ε_p^i was taken to be a constant of 1%. The values of ε_p^f were taken to be 40% for the as-received material and 12% for the irradiated material since these values corresponded to the final failure for the majority of as-received and irradiated specimens, respectively.

4.1.1 As-Received Material

The data set used for the as-received material consisted of data from all quasi-static and high strain rate tests conducted, as listed in Tables 3.1 and 3.3 respectively, with the exception of quasi-static tests conducted at 300°C and strain rates of 1×10^{-2} and $1 \times 10^{-3} \text{ s}^{-1}$ (experiments E60228A, E60228B, E60227B, E60227C, and E60227D). These tests were not used since accurate measurements of the strain were not recorded at that time. Since there were a number of tests performed at approximately the same strain rate and the stress-strain responses were very similar, the values of true-stress, true-strain rate and temperature were averaged as a function of plastic strain to provide an average response for a given strain rate. It should be noted that no further data smoothing techniques were used in any of the calculations. The complete data set for the as-received material consisted of approximately 4100 data points in a four

dimensional array (true-stress, plastic strain, true-strain rate and temperature), based on 62 experiments.

The overall effect of temperature and strain rate on the longitudinal response of the as-received material is shown in Figure 4.4. As observed with the engineering stress-engineering strain data, there is a decrease in strength with an increase in temperature, and the effect of increasing the strain rate is to increase the strength. The overall work hardening behaviour is very similar for the room temperature and elevated temperature data respectively, but the rate of work hardening is moderately higher at room temperature. This trend is consistent with results obtained for pure zirconium, where the work hardening rate increased as the temperature decreased [60].

Figure 4.5 shows the flow stress as a function of log-strain rate at various levels of plastic strain for the as-received material. As described in Section 1.3.3, the strain rate sensitivity (λ), defined as

$$(1.6) \quad \lambda = \left(\frac{\partial \sigma}{\partial \log(\dot{\epsilon}_p)} \right)_{T, \epsilon_p}$$

is greater within the high strain rate regime. This bi-linear dependence of flow stress on log-strain rate is similar to that observed for hcp α -titanium [97], but decreases significantly as the plastic strain increases from 2 to 8%.

4.1.2 Irradiated Material

The data set used for the irradiated material consisted of data from each high strain rate test listed in Tables 3.5 and 3.6, and unpublished quasi-static data obtained from 5 tests performed in a previous investigation [163]. The quasi-static tests were conducted on material from the same pressure tube as the high strain rate tests (tube P3J09). The fluence of these quasi-static specimens ranged from $6.6 - 6.9 \times 10^{25} \text{ n/m}^2$ and the nominal dimensions of the gauge length, thickness and width were 12.7, 4.2 and 4.0 mm, respectively. The quasi-static tests were conducted at a constant strain rate of $8.33 \times 10^{-3} \text{ s}^{-1}$ and temperatures ranging from room temperature to 290°C following the standards outlined in [161]. It should be noted, that the data for the irradiated material was not averaged as a function of strain rate and no data smoothing techniques were used in any of the calculations. The complete data set for the irradiated material consisted of approximately 2100 data points, based on 42 experiments.

Figure 4.6 shows the effect of temperature and strain rate on the strength of irradiated material. Comparing Figure 4.6 to Figure 4.4, it is seen that the effect of irradiation is to increase the strength for all of the conditions considered. Also, it is seen that the increase in strength due to an increase in strain rate is greater for the irradiated material. As observed for the as-received material, there is a decrease in strength with an increase in temperature. There appears to be little effect of irradiation on the work hardening behaviour at quasi-static strain rates.

Figure 4.7 shows the flow stress as a function of log-strain rate at various levels of plastic strain rate for the irradiated data. As observed with the as-received material,

the rate sensitivity is greater within the high strain rate regime. The high rate sensitivity, however, is stronger for the irradiated specimens, compared to the as-received specimens in Figure 4.5. From the data presented, it appears that the rate sensitivity for irradiated Zr-2.5Nb pressure tube material is also dependent on temperature. The rate sensitivity at 290°C is less than at room temperature, over the high strain rate regime, based on the slopes in Figure 4.7.

4.2 Constitutive Modelling of As-Received Material

4.2.1 Johnson-Cook Constitutive Relation

As described in Section 1.5.1, the Johnson-Cook constitutive relation is an empirically based model that calculates the flow stress as a function of plastic strain, strain rate and temperature by,

$$(1.22) \quad \bar{\sigma} = (A + B\bar{\epsilon}_p^n)(1 + C \ln \dot{\epsilon}^*)(1 - T_h^m)$$

From the form of equation 1.22 it is seen that the effect of increasing the strain rate will increase the strain rate term,

$$1 + C \ln \dot{\epsilon}^* \tag{4.2}$$

to values in excess of unity thereby increasing the flow stress. Increasing the temperature has the opposite effect since the thermal term,

$$1 - T_h^m \quad (4.3)$$

will decrease (< 1) thereby decreasing the flow stress. Although this approach does qualitatively account for the expected changes in flow stress as a result of changes in strain rate and temperature, it does not account for changes to the work hardening behaviour in a quantitative manner. The flow stress as a function of plastic strain response, as defined by the empirical work hardening term,

$$A + B\bar{\epsilon}_p^n \quad (4.4)$$

is simply scaled up or down, depending on the values of strain rate and temperature.

Isothermal - Constant Strain Rate Fit

In order to determine the parameters for the Johnson-Cook model, data from a minimum of three tests is required [125, 126]. Traditionally, one quasi-static test at room temperature is used to determine the parameters for the work hardening term (A, B, n), one high strain rate test to determine the parameter for the strain rate term (C), and an elevated temperature test to determine the parameter for the thermal term (m). Since it is assumed that the strain rate and temperature remain constant throughout these tests, this

procedure yields an isothermal - constant strain rate fit. This procedure was used in the current research to obtain initial parameters (isothermal parameters) for the as-received material and provide a comparison to the adiabatic - true-strain rate fit.

Three subsets of data were used for the initial fit; room temperature quasi-static data (RT, $1 \times 10^{-4} \text{ s}^{-1}$), room temperature high strain rate data (RT, 2650 s^{-1}), and elevated temperature quasi-static data (300°C , $1 \times 10^{-4} \text{ s}^{-1}$). First, the yield stress (A) and work hardening parameters (B, n) were determined by fitting the room temperature, quasi-static data to equation 4.4. The values obtained for the parameters were $A = 585 \text{ MPa}$, $B = 408 \text{ MPa}$, and $n = 0.298$. Next, room temperature, high strain rate data was fit to the product of equations 4.2 and 4.4 in order to determine the strain rate sensitivity term, C . The values for A , B , and n were fixed and it was assumed the strain rate was constant (2650 s^{-1}). The value of the reference strain rate ($\dot{\epsilon}_0$) was taken to be $1 \times 10^{-4} \text{ s}^{-1}$ since the yield stress and work hardening parameters were calculated above with data obtained at this strain rate, and the effective plastic strain rate ($\dot{\epsilon}^*$) was calculated using the nominal strain rate with equation 1.23. The resulting value for C was 1.00×10^{-2} . Finally, fixing the above values for A , B , n , and C , the thermal exponent, m , was determined by fitting elevated temperature, quasi-static data to equation 1.22. The resulting value for m was 0.698. For this calculation, the homologous temperature (T_h), as defined in equation 1.24, was calculated assuming $T_{room} = 293 \text{ K}$, $T_{melt} = 2033 \text{ K}$, and the specimen temperature (T) was constant at 573 K . These parameters obtained for the isothermal - constant strain rate fit (isothermal parameters) are summarised in Table 4.1.

The calculated stress-strain curves from the Johnson-Cook isothermal - constant strain fit to the as-received data are compared to the experimental data in Figure 4.8. It is

seen that the predicted quasi-static values match the experimental data very well. The high strain rate data, however, is significantly under-estimated by the model. At 2% plastic strain, the predicted stress lies approximately 60 and 125 MPa below the measured values for the room temperature and 290°C tests, respectively. The difference between the predicted and actual stress increases as the plastic strain value increases. This error exists since the initial fitting process did not consider the increase in strain rate and temperature for the high strain rate data over the duration of each test.

Adiabatic - True-Strain Rate Fit

The isothermal parameters were then used as starting values for a non-linear regression fit to the complete data set for the as-received material that accounted for increases in strain rate and temperature during the high strain rate tests. The resulting parameters, denoted as the adiabatic parameters, and the associated uncertainties are also summarised in Table 4.1. It should be noted that a variety of starting values for the parameters were considered and in all cases the regression routine converged to the same values listed in Table 4.1. Comparing the initial and adiabatic parameters, it is seen that the yield stress term (A) decreased. The work hardening terms (B and n) also changed resulting in a net increase in the degree of work hardening, for a given value of plastic strain. The strain rate and thermal terms (C and m) both increased relative to the values obtained in the isothermal - constant strain rate fit, indicating that the adiabatic - true-strain rate fit better captures the rate and temperature dependency.

Figure 4.9 provides comparisons of the predicted (adiabatic) and experimental stress-strain curves. It is seen that the agreement between the experimental data and the response calculated by the adiabatic Johnson-Cook fit is much better than the isothermal fit (Figure 4.8). The model, however, slightly under-predicts the stress at room temperature and quasi-static rates for all values of plastic strain, and at elevated temperatures for strains less than approximately 10% under high strain rate loading conditions.

4.2.2 Zerilli-Armstrong hcp Constitutive Relation

The data obtained from testing of the as-received material was also used to determine the parameters (σ_a , B , B_0 , β , and α) for the Zerilli-Armstrong hcp constitutive model, given by

$$(1.33) \quad \bar{\sigma} = \sigma_a + B e^{-\beta \tau} + B_0 \epsilon^{1/2} e^{-\alpha \tau}$$

and described in Section 1.5.2. The method used by Zerilli and Armstrong [130, 165] to determine the parameters assumes that the temperature and strain rate remain constant throughout the test. This method was implemented in the current research to obtain an initial fit, and again was denoted as the isothermal - constant strain rate fit. The resulting parameters were then applied as starting values for a non-linear regression on the complete data set, which included the variations in temperature and strain rate. These parameters were denoted as the adiabatic - true strain rate parameters.

Isothermal - Constant Strain Rate Fit

The method used by Zerilli and Armstrong [130, 165], is based on the assumption that the flow stress ($\bar{\sigma}$) has a linear dependence on the square root of the plastic strain represented by,

$$\bar{\sigma} = \sigma_y + K\varepsilon_p^{1/2} \quad (4.5)$$

where the yield stress (σ_y) and strain hardening coefficient (K) are defined as,

$$\sigma_y = \sigma_a + Be^{-\beta\tau} \quad (4.6)$$

and

$$K = B_0e^{-\alpha\tau} \quad (4.7)$$

To obtain values for σ_y and K , straight lines are fit to true-stress as a function of the square root of plastic strain using a linear regression routine. The values of σ_y and K are then tabulated with the corresponding average strain rate (assumed constant) and test temperature (assumed constant) for each test performed. The Zerilli-Armstrong hcp parameters σ_a , B , β_0 , and β_1 are then obtained by fitting equation 4.6 to the yield stress,

strain rate and temperature data, and the parameters B_0, α_0 , and α_1 are obtained by fitting equation 4.7 to the hardening coefficient, strain rate and temperature data. To make this process more convenient, equations 4.6 and 4.7 are written respectively as,

$$\sigma_y = \sigma_a + B e^{-B \xi_\beta} \quad (4.8)$$

and

$$\ln K = \ln B_0 - \alpha_1 \xi_\alpha \quad (4.9)$$

where

$$\xi_\beta = T \ln \left(\frac{\dot{\epsilon}_{0\beta}}{\dot{\epsilon}} \right) \quad (4.10)$$

and

$$\xi_\alpha = T \ln \left(\frac{\dot{\epsilon}_{0\alpha}}{\dot{\epsilon}} \right) \quad (4.11)$$

Then, plotting the yield stress as a function of ξ_β , with a suitable choice of $\dot{\epsilon}_{0\beta}$, all the points should lie on a single curve. Similarly, with the proper choice of $\dot{\epsilon}_{0\alpha}$ all the points

should lie on a single line when plotting the logarithm of the hardening coefficient as a function of $\dot{\xi}_\alpha$. The parameters α and β are defined as [130, 165],

$$\alpha = \alpha_1 \ln \left(\frac{\dot{\varepsilon}_{0\alpha}}{\dot{\varepsilon}} \right) \quad (4.12)$$

and

$$\beta = \beta_1 \ln \left(\frac{\dot{\varepsilon}_{0\beta}}{\dot{\varepsilon}} \right) \quad (4.13)$$

where

$$\alpha_0 = \alpha_1 \ln(\dot{\varepsilon}_{0\alpha}) \quad (4.14)$$

and

$$\beta_0 = \beta_1 \ln(\dot{\varepsilon}_{0\beta}) \quad (4.15)$$

This approach was adopted by Zerilli [165] to simplify the fitting process and minimise the need for iterative computer calculations.

The true-stress as a function of square root of plastic strain was calculated for the entire as-received data set and used to determine the linear coefficients for equation 4.5, corresponding the slope (K), which represents the hardening coefficient, and the y-intercept (σ_y), which represents the yield stress. Figure 4.10 shows the resulting linear fits to the true-stress versus square root plastic strain data for four tests, and Table 4.3 contains all of the values obtained for the as-received material. This data provides the basis to obtain an initial fit to the Zerilli-Armstrong hcp constitutive model, as described by equation 1.33. From Figure 4.10, it is seen that both σ_y and K depend on strain rate and temperature. Based on this observation and the theoretical basis of the Zerilli-Armstrong constitutive models, as described in Section 1.5.2, it is anticipated that the constitutive behaviour of the as-received material will exhibit typical hcp characteristics. That is, an intermediate behaviour having characteristics of both bcc and fcc typical behaviour. For bcc materials, K is generally independent of strain rate and temperature implying that $\alpha_0 = \alpha_1 \equiv 0$, while for fcc materials, σ_y is independent of strain rate and temperature implying that $\beta_0 = \beta_1 \equiv 0$.

The values of ξ_α and ξ_β were calculated using equations 4.11 and 4.10, respectively, assuming $\dot{\epsilon}_{0\alpha} = 1 \times 10^6 \text{ s}^{-1}$ and $\dot{\epsilon}_{0\beta} = 1 \times 10^9 \text{ s}^{-1}$. Figure 4.11 shows the yield stress as a function of ξ_β and the resulting fit of equation 4.8, which yielded values for σ_a , B and β_1 of 386 MPa, 897 MPa, and $1.40 \times 10^{-4} \text{ K}^{-1}$, respectively. Then using equation 4.15 β_0 was calculated to be $2.90 \times 10^{-3} \text{ K}^{-1}$. The natural logarithm of the hardening coefficient was then plotted as a function of ξ_α using the relationship

described in equation 4.9, and shown in Figure 4.12. Values for B_0 and α_1 were then determined to be 570 MPa and $6.01 \times 10^{-5} \text{ K}^{-1}$, respectively. In keeping with the method used by Zerilli and Armstrong, the high strain rate data was not used to calculate B_0 and α_1 since the temperature of the specimen increases as a result of adiabatic heating [129, 130]. From equation 4.14, α_0 was then calculated to be $8.30 \times 10^{-4} \text{ K}^{-1}$. The values from the isothermal - constant strain rate fitting procedure and the estimated uncertainty are provided in Table 4.4.

A comparison of the predictions from the isothermal - constant strain rate fit with the measured data is shown in Figure 4.13. It is seen that the agreement is good for the quasi-static data, similar to the predicted response using the isothermal - constant strain rate Johnson-Cook fit (Figure 4.8). Unlike the Johnson-Cook model, however, the Zerilli-Armstrong model over-predicts the strength and work hardening rate for the high strain rate data. The difference between the high strain rate experimental data and the Zerilli-Armstrong predictions increases for an increase in plastic strain. This discrepancy is associated with thermal softening as a result of adiabatic heating [130, 165], which has not been considered in the isothermal - constant strain rate fitting procedure.

Adiabatic - True-Strain Rate Fit

The isothermal parameters were then used as starting estimates for an adiabatic fit of regression on all the parameters. The resulting values for $\sigma_a, B, \beta_0, \beta_1, B_0, \alpha_0$, and α_1 were 443 MPa, 787 MPa, $3.46 \times 10^{-3} \text{ K}^{-1}$, $1.59 \times 10^{-4} \text{ K}^{-1}$, 766 MPa, $1.84 \times 10^{-3} \text{ K}^{-1}$, and $2.01 \times 10^{-5} \text{ K}^{-1}$, respectively. These values and the estimated uncertainties are listed in

Table 4.4. A comparison of the predicted stress-strain curves with experimental data is shown in Figure 4.14. From the figure it is seen that the predictions determined from the Zerilli-Armstrong adiabatic fit are in very good agreement with the measured data. The difference between the initial (Figure 4.13) and adiabatic (Figure 4.14) fits is similar to that reported by Zerilli and Armstrong when they applied their hcp model to the titanium alloy, Ti-6Al-4V [130]. In that investigation, it was stated that in order to accurately predict the stress-strain behaviour of high strain rate deformation, thermal softening due to adiabatic heating had to be considered [130]. The results from the current research further demonstrate the importance of including the effects of adiabatic heating when modelling material behaviour under high strain rate conditions. Comparing these results to the Johnson-Cook adiabatic fit, shown in Figure 4.9, it is seen that the Zerilli-Armstrong model is slightly more accurate over the strain rate and temperature range considered.

Generally, the effect of temperature on work hardening rate for high strain rate deformation is small for low values of strain, but increases significantly for large strains. To account for this type of behaviour in bcc materials, Goldthorpe [142] added a term to the Zerilli-Armstrong bcc model. The method implemented in the current research, however, not only takes into account the effect of adiabatic heating, thereby making the model applicable over a larger range of strains, but it also accounts for the increase in strain rate that occurs in high strain rate deformation. It is expected that this approach will improve the predictive capability of the Zerilli-Armstrong hcp constitutive model when used in numerical simulations of events involving dynamic loading conditions.

4.3 Constitutive Modelling of Irradiated Material

Based on the constitutive modelling of the as-received material, it was apparent that the adiabatic - true strain rate fits produced the most accurate results. Therefore, only adiabatic fits were performed for the irradiated material. All of the available data for the irradiated material, as described in Section 4.1.2, was used to determine adiabatic - true-strain rate parameters for both the Johnson-cook and Zerilli-Armstrong hcp constitutive relations. Again, optimisation of the parameters to three significant figures was accomplished through use of the non-linear least squares regression routine [164].

4.3.1 Johnson-Cook Constitutive Relation

The calculated values for A , B , n , C , and m , from the regression of the Johnson-Cook relation were 835 MPa, 691 MPa, 0.472, 1.10×10^{-2} , and 0.728, respectively. These values and the associated uncertainties are given in Table 4.2. It should be noted that the value of the reference strain rate ($\dot{\epsilon}_0$) for the irradiated material was taken to be $8.3 \times 10^{-4} \text{ s}^{-1}$ since the quasi-static data was obtained at this strain rate. Comparing the uncertainties in the parameters for the irradiated fit to those obtained for the as-received adiabatic fit, it is seen that the uncertainties associated with the work hardening terms (B , and n) for the irradiated material are larger. This result indicates that there is a large variability in the work hardening rate for the irradiated material, which is

consistent with the observed stress-strain behaviour shown for the irradiated material in Figure 4.6 and Appendix F.

Figure 4.15 compares the experimental stress-strain curves to those predicted by the Johnson-Cook adiabatic fit for the irradiated material. It is seen that the adiabatic fit predicts the response of the irradiated material at quasi-static rates reasonably well, but predictions of the high strain rate data are not as accurate. It is noted that the measured work hardening behaviours at strain rates of 500 s^{-1} and 2300 s^{-1} are quite different. The corresponding constitutive fits are close since the difference in rates is relatively small. Consequently, uncertainties on the order of 200 MPa are observed, largely attributed to the poor repeatability of the high strain rate tests on irradiated material. Since there is less variation in the work hardening behaviour of the irradiated material at high strain rates and 290°C , the predictions at 290°C are more accurate.

4.3.2 Zerilli-Armstrong hcp Constitutive Relation

The values for $\sigma_a, B, \beta_0, \beta_1, B_0, \alpha_0$, and α_1 resulting from the regression of the Zerilli-Armstrong hcp relation were 644 MPa, 628 MPa, $3.60 \times 10^{-3} \text{ K}^{-1}$, 0 K^{-1} , 3242 MPa, $4.70 \times 10^{-3} \text{ K}^{-1}$, and $2.06 \times 10^{-5} \text{ K}^{-1}$, respectively. The value of β_1 decreased to values less than 1×10^{-10} with successive iterations of the regression routine, so it was set to zero. The Zerilli-Armstrong hcp adiabatic parameters and the associated uncertainties for the irradiated material are provided in Table 4.5.

Comparing the values of the adiabatic parameters obtained for the as-received and irradiated material, it is seen that the observed increase in strength due to irradiation is

reflected in an increase to the athermal stress parameter (σ_a). The changes in the other terms that contribute to the yield stress behaviour (B , β_0 and β_1) indicate that the yield stress of the irradiated material is less strain rate and temperature dependent compared to the as-received material. This suggests that the irradiated material behaves more like an fcc material, which is further supported by the increase in the hardening coefficient terms B_0 , α_0 and α_1 . As seen with the Johnson-Cook adiabatic fit, the estimated uncertainties for the irradiated material are significantly larger than those calculated for the as-received material. Again, these discrepancies are a result of the large variation in the work hardening behaviour of the irradiated material at high strain rates.

Figure 4.16 shows a comparison of the measured and predicted stress-strain response calculated by the Zerilli-Armstrong hcp model using the derived adiabatic parameters. Similar to the comparison of the Johnson-Cook adiabatic fits to the irradiated data (Figure 4.15), the predicted quasi-static response is more accurate than the high strain rate response, especially at room temperature. The Zerilli-Armstrong adiabatic predictions, however, appear to be slightly better than the Johnson-Cook adiabatic predictions. The largest difference in the predictions of the Zerilli-Armstrong model, as compared to the Johnson-Cook model, is seen in the change in predicted work hardening behaviour as the strain rate is increased from quasi-static to high strain rate conditions. This is reflected in the observed change in slope of the predictions for the quasi-static and high strain rate tests in Figure 4.16. The Johnson-Cook predictions in Figure 4.15 remain basically parallel for the range of strains considered. Also, the flow stress predicted by the Zerilli-Armstrong model, for an increase in strain rate from 500 to 2100 s⁻¹, is closer to the measured data than that predicted by the Johnson-Cook model.

The Zerilli-Armstrong adiabatic fit, however, does not accurately predict the room temperature work hardening behaviour at the highest strain rates (2100 s^{-1}). This is also a reflection of the large variation in the measured work hardening behaviour for the irradiated material. However, by using all of the possible data and not averaging the measured responses, the process of calculating the parameters for the constitutive model provides a superior fit to the experimental data.

Chapter 5

Numerical Simulations

Numerical simulations of the direct-tension split Hopkinson bar experiments were performed to verify that the constitutive parameters derived in Chapter 5 could be used to model the longitudinal deformation of as-received and irradiated Zr-2.5Nb pressure tube material within a finite element formulation. Both the Johnson-Cook and Zerilli-Armstrong hcp constitutive relations were used to simulate the material behaviour. In these simulations, only the specimen geometry was considered, but strain-time histories from the experiments performed in this investigation were used to determine the boundary conditions for the model.

The numerical simulations were performed with the three-dimensional finite element code, LS-DYNA [166], developed by Livermore Software Technology Corporation (LSTC). LS-DYNA is a lumped-mass, Lagrangian code that uses the explicit central difference method to integrate the equations-of-motion with respect to time [8, 9, 166]. LS-DYNA allows for the implementation of additional constitutive relations by way of user-defined subroutines that can be compiled as part of the

executable code. Currently, LS-DYNA does not offer the Zerilli-Armstrong hcp constitutive relation, so it was implemented as a user-defined subroutine as part of the current research. Single-element calculations were performed to verify that the constitutive relation was implemented correctly as well as detailed modelling of the entire specimen geometry.

Details of the implementation and verification of the Zerilli-Armstrong hcp constitutive relation in LS-DYNA are provided in Section 5.1. The finite element model and the boundary conditions used in the numerical simulations of the direct-tension split Hopkinson experiments are presented in Section 5.2, and the results from the simulations are described in Section 5.3.

5.1 Constitutive Model Implementation and Verification

In a previous investigation [167], a user-defined subroutine, *UMAT46*, was written that calls a second subroutine, *GFLOW*, used to calculate the flow-stress based on a number of constitutive relations. In the current investigation, the Zerilli-Armstrong hcp constitutive relation was incorporated into the existing version of *GFLOW*. It should be noted that the Johnson-Cook constitutive relation had previously been incorporated [167] into *GFLOW*, and therefore, did not have to be added as part of this research.

To verify that the Zerilli-Armstrong hcp and the Johnson-Cook constitutive relations in the previous investigation [167] had been implemented correctly, single-element finite element calculations were performed with an executable version of LS-DYNA that had been compiled with the modified version of *GFLOW*. For these

calculations a single eight-node solid element with single-point integration was used. One end of the element was held fixed and the opposite end was displaced at a prescribed velocity until the effective plastic strain had reached a value of unity. Three different cases were considered and included a room temperature - low strain rate case, a room temperature - high strain rate case, and a 300°C - low strain rate case. Room temperature (RT) was taken to be 20°C. The low strain rate cases were conducted at a strain rate of 1 s^{-1} , and the high strain rate case was conducted at a strain rate of 1000 s^{-1} . Separate calculations were performed using both the Zerilli-Armstrong hcp and Johnson-Cook constitutive relations to describe the material behaviour and the adiabatic parameters for the as-received material, as outlined in Tables 4.1 and 4.4. The calculated flow stress curves as a function of effective plastic strain are shown in Figure 5.1 for all of the cases. Included in Figure 5.1 are the corresponding analytically calculated stress-strain curves, which were obtained by taking the corresponding values of effective plastic strain, strain rate and temperature, and applying the Johnson-Cook (equation 1.22) and Zerilli-Armstrong hcp (equation 1.33) relations. It is seen that the single-element and analytical calculations are in close agreement for all of the conditions considered. These results indicate that the equations for the constitutive relations were implemented correctly into the *GFLOW* subroutine.

5.2 Finite Element Model and Boundary Conditions

The numerical simulations of the direct-tension split Hopkinson bar experiments performed in this investigation only considered the geometry of the

specimen and not any portion of the Hopkinson bar apparatus. This method of modelling Hopkinson bar experiments has recently been used in a number of investigations for tensile [153, 162, 167, 168, 100, 102] and torsional [169] specimens, and is generally accepted to produce accurate results if the appropriate boundary conditions are applied. Since there is symmetry in the x-y plane of the specimen, only one half of the specimen was modelled, as indicated in Figure 5.2, which shows the adopted finite element mesh. The model consisted of 10,854 nodes and 8,330 eight-node hexahedron solid continuum elements. The same discretization was used for all of the calculations. The gauge length region in the model is shown in Figure 5.3 and was comprised of six elements through the width (x-direction), twenty elements along the length (y-direction), and five elements through the thickness (z-direction). Since the model represents a longitudinal specimen, the x, y, and z-directions indicated in Figure 5.3 correspond to the transverse, longitudinal and radial directions in a pressure tube, respectively. The finite element model was developed using the LSTC mesh generation software, INGRID [170]. The sensitivity of the results on the mesh was investigated in preliminary calculations. Although the details of those calculations are not presented here, it was determined that further refinement of the mesh shown in Figure 5.2 did not result in any change to the magnitude or distribution of stress in the gauge length region.

The material response of the as-received and irradiated longitudinal Zr-2.5Nb specimens was modelled using both the Johnson-Cook and Zerilli-Armstrong (hcp) constitutive relations. The adiabatic parameters for the Johnson-Cook and Zerilli-Armstrong models, described in Sections 4.2 and 4.3, were used for all of the simulations.

To simulate the conditions in a split Hopkinson bar experiment, one end of the specimen model was held fixed (bottom) and the other end was displaced (top). The fixed end was constrained by restricting displacement of all nodes at the bottom to the x-z plane. The displacement of the other end was imposed as a prescribed velocity-time history in the y-direction to all the nodes at the top of the model. The specific velocity-time histories were determined by calculating the difference in the velocities of the ends of the specimen (V_{diff}) from experimental data by,

$$V_{diff} = E_{bar} / \rho_{bar} (2\varepsilon_i - \varepsilon_t) \quad (5.1)$$

where E_{bar} and ρ_{bar} are the elastic modulus and density of the Hopkinson bar, respectively. Also, ε_i and ε_t are the strain-histories from the incident and transmitted strain gauges, respectively. Figure 5.4 shows an example of a calculated velocity-time history used in the simulations. This velocity profile corresponds to an experiment (E60223D) conducted on as-received material at an average strain rate of 1090 s^{-1} and 290°C . For all of the numerical simulations performed in this investigation, raw experimental data was used to calculate the velocity profile. In [162] it was shown that this technique of applying the required displacement as a boundary condition produces accurate results.

Four separate split Hopkinson bar experiments were simulated in this investigation using both the Johnson-Cook and Zerilli-Armstrong (hcp) constitutive relations to model the material behaviour. A room temperature and a 290°C experiment for both as-received and irradiated material were simulated. The experiments performed

on as-received material that were simulated are tests E50327A (RT, strain rate = 1060 s^{-1}) and E60223D (290°C, strain rate = 1090 s^{-1}). For the irradiated material, tests E70610A (RT, strain rate = 1360 s^{-1}) and E80310D (290°C, strain rate = 2010 s^{-1}) were simulated.

5.3 Numerical Results

The resultant force along the bottom (fixed-end) of the model was calculated as a function of time for each simulation and written to a file. These force-histories represent the forces that would be transmitted to the output bar (F_t) in an experiment, and were converted to specimen stress-histories ($\sigma_s(t)$) using,

$$\sigma_s(t) = \frac{F_t}{A_s} \quad (5.2)$$

where A_s is the original cross-sectional area of the specimen. Then, using

$$\varepsilon_s(t) = \frac{\Delta L}{L_0} \quad (5.3)$$

to calculate the strain-history of the specimen, a predicted engineering stress-strain curve was obtained.

5.3.1 As-Received Material

Comparisons of the predicted stress-strain curves to measured values from experiments performed on as-received material are shown in Figure 5.5. It should be noted that no data smoothing techniques were used in these calculations, and the observed oscillation in the predicted response is primarily a result of modelling only the specimen geometry. In the experiments, the incident stress wave is partially transmitted through the specimen into the output bar, denoted as the transmitted wave. In the numerical simulations, the bottom end of the specimen is fixed and the entire transmitted wave is reflected back into the specimen causing the oscillation. From Figure 5.5 it is seen that predictions using both constitutive relations for the as-received material compare very well with the measured values. The predicted curves accurately capture the stress-strain behaviour to final failure, which demonstrates that the constitutive relations are capable of modelling the thermal-softening behaviour at high strain rates. From Figure 5.5 it is also seen that the results from both material models are very similar. This result was expected, since analytical predictions from both constitutive relations compared closely to measured data for the as-received material when adiabatic parameters were used, as shown in Figures 4.9 and 4.14.

5.3.2 Irradiated Material

Comparisons of the predicted and measured stress-strain curves for the irradiated material are shown in Figure 5.6. It is seen that the stress-strain behaviour of the

irradiated material was not as accurately predicted as for the as-received material. The overall effect of strain rate and temperature on strength, however, was captured with reasonable accuracy. Again, both the Johnson-Cook and Zerilli-Armstrong hcp relations produced very similar results. As discussed in Section 4.3, the constitutive parameters derived for the irradiated material have a larger uncertainty since the stress-strain behaviour of the irradiated material was less repeatable. As a result, the finite element predictions are expected only to capture the overall effects of strain rate and temperature.

Comparing the numerical results of the as-received and irradiated material presented in Figures 5.5 and 5.6, it is seen that the oscillations for the predictions of the irradiated material behaviour are larger. The increase in the magnitude of the oscillation is a result of the higher strain rates considered for the irradiated cases, since the incident and transmitted forces are proportional to strain rate. The oscillation in the simulations using the Zerilli-Armstrong constitutive relationship decrease as the strain increases. The oscillations in the Johnson-Cook predictions, however, do not appear to decrease. Also, the thermal-softening behaviour exhibited in the room temperature test is more closely approximated by the Zerilli-Armstrong hcp model. These results further indicate that the Zerilli-Armstrong hcp constitutive relationship should produce more accurate results when used to model pressure tube rupture scenarios.

5.3.3 Deformation Behaviour

The ability of the finite element method used in the current research to capture the deformation behaviour of Zr-2.5Nb pressure tube material under high strain rate loading

was investigated by comparing the predicted deformation in the gauge length region to experimental data. In all of the simulations performed, localised deformation (necking) was observed to occur near the center of the gauge length, as shown in Figures 5.7 and 5.8. This observation is consistent with experimental observations. The deformed mesh shown in the figures is from a numerical simulation of an experiment (E50404C) performed at room temperature and a average strain rate of 1580 s^{-1} . The Zerilli-Armstrong hcp constitutive relationship was used to model the material behaviour. The accumulated engineering strain for the deformed mesh shown in Figure 5.7 is 19% and in Figure 5.8 it is 25%. Results from experiment E50404C (Appendix D1) indicate that the specimen failed at approximately 25% engineering strain, which is the same level of strain represented in Figure 5.8. Results from the finite element calculations of experiment E50404C at this point showed that the effective plastic strain was between 50% and 60% in the neck region. Based on these results, it can be assumed that as-received Zr-2.5Nb pressure tube material tested at high strain rates and room temperature fails when the effective plastic strain reaches approximately 50%.

Results from the numerical simulations were used to calculate the reduction in cross-sectional area as a function of engineering strain. This was performed by multiplying the thickness (longitudinal-radial plane) and width (longitudinal-transverse plane) across the gauge center. Predicted reduction in cross-sectional area values were calculated at 5%, 10%, 19%, and 25% engineering strain and are shown in Figure 5.9. Included in the figure are measurements from a number of experiments on as-received pressure tube material, as described in Section 2.6.1. It is seen that the predicted values

are in good agreement with the experimental data indicating that the finite element method used is able to capture the observed localised deformation to the point of failure.

The predicted deformation at 19% engineering strain is compared in Figure 5.10 to the observed deformation for a specimen tested at approximately 1700 s^{-1} using the split Hopkinson bar at Carleton University. As discussed in Sections 2.6.1 and 3.3, the specimen shown in Figure 5.10 was deformed to 19% engineering strain and then the test was interrupted. From the figure it is seen that the finite element model under-predicts the amount of strain in the radial direction (Figure 5.10a) and over-predicts the amount of strain in the transverse direction (Figure 5.10b). The predicted radial and transverse strains were 15.0% and 13.7%, respectively, and the measured radial and transverse strains were 20.2% and 2.0% respectively. This discrepancy is due to the anisotropic behaviour of Zr-2.5Nb pressure tube material, which was not considered in the development of the material model or the calculations. It is interesting to note that the under-prediction is in the direction oriented favourably for slip (radial direction) and the over-prediction is in the direction where deformation is primarily by twinning (transverse direction). This demonstrates that the current finite element approach is not capable of predicting the directional dependent deformation characteristics of Zr-2.5Nb pressure tube material. The overall reduction in area predictions, however, are close to the measured values and it is anticipated that the expected increase in stress, strain rate and temperature with necking will be captured with reasonable accuracy.

Chapter 6

Discussion

The measured mechanical behaviour of the as-received and irradiated Zr-2.5Nb pressure tube material is discussed in Section 6.1. The constitutive modelling of this data and the numerical simulations of the experiments are discussed in Section 6.2.

6.1 Mechanical Behaviour

Results from the experiments have revealed similar characteristics in the stress-strain behaviour of as-received and irradiated Zr-2.5Nb pressure tube material, as a function of tube orientation, for strain rates ranging from 1×10^{-4} to 3000 s^{-1} , temperatures from 20 to 300°C , and irradiation fluences up to $8 \times 10^{25} \text{ n/m}^2$. The pressure tube material in the longitudinal orientation exhibited a moderate, but steady amount of work hardening until final failure, which is consistent with slip-controlled deformation [74]. The rate of work hardening was seen to increase for increasing strain rate and decreasing temperature, similar to results on pure zirconium [59, 60]. In the

transverse direction, the pressure tube material did not strain harden gradually, but rather increased in strength rapidly to the point of σ_{UTS} . Beyond this point, the material strain softened rapidly until final failure. This behaviour is also consistent with previous work [74] and is attributed to the transverse specimens being favourably oriented for deformation twinning. As a result of this anisotropy, Zr-2.5Nb pressure tube material was stronger in the transverse direction, as seen in previous quasi-static tests [74, 81]. Over the high strain rate range considered in this research, the as-received transverse specimens were approximately 10% to 30% stronger and the irradiated transverse specimens were 12% to 20% stronger.

The main effect of irradiation on the mechanical behaviour was to increase the strength for all of the conditions considered. The observed increase in strength at the quasi-static rates was consistent with previous work on pressure tubes removed from operating reactors [73, 74, 81]. The irradiation strengthening is due to an increase in the dislocation density and the number of dislocation pinning sites caused by the formation of Frenkel pairs upon irradiation [13, 16, 67]. Although there was significant strengthening due to irradiation, the strength did not increase noticeably as the fluence increased above approximately 1 n/m^2 . This behaviour is consistent with observations in [73] and [81], and indicates that the majority of the irradiation induced strengthening in Zr-2.5Nb pressure tube material occurs at a fluence level $\leq 1 \text{ n/m}^2$.

A second effect of irradiation on the mechanical behaviour over the high strain rate regime was to decrease the ductility for both specimen orientations. As a result of irradiation, the failure strain decreased by 7% and 10% strain for the longitudinal specimens at room temperature and 290°C, respectively. For the transverse specimens,

the reduction in failure strain due to irradiation was approximately 3% strain over the temperature range considered. It should be noted, however, that all of the irradiated specimens tested in this investigation failed in a ductile manner and a significant amount of plastic strain was observed before final failure.

Compared to the high strain rate data obtained in this research for the as-received material, there was a large variability in the irradiated data. This scatter was attributed to the difficulties associated with remotely machining the irradiated specimens and attaching the specimens to the adapters. The variability in the data, therefore, could be decreased if the machining techniques could be improved, or the method of attaching the specimen to the split Hopkinson bar was changed. In [118], the specimens were glued to the Hopkinson bar, thereby eliminating the need for a mechanical connection. This technique was not applied in the current research due to the design limitations associated with machining specimens from thin-walled pressure tubes. However, a miniaturised specimen design, as used to by other investigators to test materials at high rates in compression [9], could be attempted for Zr-2.5Nb pressure tube material.

The main effect of increasing the rate of strain, or decreasing the temperature, for both the as-received and irradiated specimens was to increase the strength over all of the conditions considered. A bi-linear dependence of strength on log-strain rate existed for as-received and irradiated material in both orientations, with the strain rate sensitivity being greater over the high strain rate regime. This bi-linear dependency did not change with temperature. An increase in strength as a result of increasing the strain rate is usually attributed to structure evolution, which includes such phenomena as grain boundary interactions, and dislocation generation, interaction and movement. Over the

quasi-static strain rate range there was a moderate increase in strength for both the longitudinal and transverse orientations, and it is speculated that the observed increase in strength was due to rate sensitivity of structure evolution. Although this strengthening mechanism is commonly used to explain an increase in strength at high strain rates, strengthening can also be caused by a change in the mode of deformation. In this investigation, the strength and initial strain hardening behaviour changed continuously for as-received transverse specimens as the strain rate was increased. A distinct yield drop occurred and increased in magnitude over the high strain rate range considered. Similar strengthening behaviour was observed in other studies on zirconium and was attributed to an increase in the amount of deformation twinning [59, 60]. Although a detailed examination of the substructure evolution was not undertaken in the current investigation, it is speculated that increasing the strain rate had a similar effect on Zr-2.5Nb pressure tube material and that the propensity of twinning increased for the transverse specimens over the high strain rate regime.

6.2 Constitutive Behaviour and Numerical Simulations

The data base acquired in the experimental component of this research provides a basis for the development of physically-based constitutive relations capable of predicting the deformation behaviour of as-received and irradiated Zr-2.5Nb pressure tube material. As a first step, the constitutive parameters for the Johnson-Cook [125] and Zerilli-Armstrong hcp [130] constitutive relations were determined for the longitudinal deformation of both as-received and irradiated material.

In order to accurately predict the measured stress-strain response at high strain rates, the effects of adiabatic heating and localised necking had to be considered. Accounting for these effects, both the Johnson-Cook and Zerilli-Armstrong hcp constitutive relations very accurately predicted the behaviour of the as-received material for the range of strain rates and temperatures considered. The behaviour of the irradiated material was not as closely predicted, but the overall effect of strain rate and temperature on the stress-strain behaviour was captured with reasonable accuracy. This discrepancy is attributed to the relatively large variability in the measured stress-strain responses for the irradiated material, as noted above. Overall, the predictions using the Zerilli-Armstrong hcp constitutive relation were only slightly more accurate than those produced by the Johnson-Cook relation for the conditions considered. However, it is anticipated that since the Zerilli-Armstrong hcp relation is based on the physics of dislocation interactions, it will more accurately predict the stress-strain behaviour over a wider range of conditions. This assumption is supported by examining the difference in the Zerilli-Armstrong hcp constitutive parameters derived for the as-received and irradiated material. It appears that the relation captured some of the physical changes as a result of irradiation damage. In particular, the increase to the athermal stress parameter (σ_a), as well as changes to the other terms that contribute to the yield stress behaviour (B , β_0 and β_1), indicate that the yield stress of the irradiated material is less strain rate and temperature dependent. This trend suggests that the irradiated material behaves principally like an fcc material and that the deformation behaviour is influenced more by the intersection of dislocations [127]. This explanation is reasonable since the

dislocation density, and therefore, the number of dislocation intersections will increase with irradiation.

The numerical simulations of the direct-tension split Hopkinson bar experiments performed in this investigation indicated that both the Johnson-Cook and Zerilli-Armstrong hcp constitutive relations are capable of predicting the stress-strain behaviour of Zr-2.5Nb pressure tube material in the longitudinal direction. This includes predictions of the onset of localised deformation and the reduction in cross-sectional area to the point of final failure. Closer examination of the results, however, demonstrate that the anisotropic deformation exhibited by the radial and transverse strains is not captured.

The results of the constitutive modelling and numerical simulations compare very well to experimental results for as-received material, and reasonably well to the irradiated material. However, it should be emphasised that these capabilities are merely a starting point in the development of a fully predictive model for pressure tube rupture scenarios. As indicated in previous investigations [11, 12], the longitudinal deformation of Zr-2.5Nb pressure tubes during rupture is an important parameter. However, the scenario of pressure tube rupture also involves high rate deformation in the transverse direction and unstable crack propagation. Also, based on the data obtained in this research, Zr-2.5Nb pressure tube material remains highly anisotropic at high rates and deformation involves both slip and twinning. Therefore, the development of a fully predictive model would have to incorporate material anisotropy, deformation models for both slip and twinning, and some form of a damage / failure model capable of modelling dynamic fracture.

Chapter 7

Conclusions and Recommendations

The primary objectives of the current research were: (i) to experimentally determine the mechanical behaviour of Zr-2.5Nb pressure tube material at high rates of strain; and, (ii) to characterise the material behaviour with a constitutive relation that is compatible with current analysis efforts investigating the hypothetical accident scenario of pressure tube rupture and subsequent calandria tube failure. As a result of this research, the following contributions were made:

- i) The successful application of a direct-tension split Hopkinson apparatus to determine the mechanical properties of as-received and irradiated Zr-2.5Nb pressure tube material in both the longitudinal and transverse tube orientations.
- ii) Determination of the Johnson-Cook and Zerilli-Armstrong hcp constitutive parameters for the as-received and irradiated Zr-2.5Nb pressure tube material in the longitudinal direction.

- iii) Implementation of the Zerilli-Armstrong hcp constitutive relation into an explicit, dynamic finite element code capable of modelling the deformation of materials under high strain rate loading conditions.
- iv) Numerical simulations of the direct-tension split Hopkinson bar experiments.

7.1 Conclusions

The following major conclusions stem from the current research:

- i) Zr-2.5Nb pressure tube material remains anisotropic over the strain rate, temperature and fluence ranges investigated, with the pressure tube being stronger in the transverse orientation.
- ii) The effect of irradiation is to strengthen Zr-2.5Nb pressure tube material over the high strain rate regime considered. It also appears that irradiation moderately reduces the materials ductility
- iii) A bi-linear dependence of strength on log-strain rate was observed for both as-received and irradiated material, with the strength increasing for increasing strain rate.
- iv) The irradiated material sustained a significant amount of plastic deformation over the range of strain rates considered and final failure occurred in a ductile manner.
- v) The Zerilli-Armstrong hcp and Johnson-Cook constitutive relations with the derived adiabatic parameters accurately predict the stress-strain behaviour of

as-received and irradiated Zr-2.5Nb pressure tube material in the longitudinal direction.

- vi) Finite element analysis of the direct-tension split Hopkinson bar experiments using the derived constitutive parameters accurately predicted the stress-strain behaviour of the specimens. The onset of necking and reduction in cross-sectional area were captured reasonably well, however, the model did not capture the observed anisotropy.

7.2 Recommendations

The following recommendations are made based on the findings of the current research:

- i) The constitutive behaviour of Zr-2.5Nb pressure tube material should be characterised in the transverse direction with a relation capable of modelling the physics of deformation twinning.
- ii) The development of an anisotropic constitutive relation, incorporating the physics associated with both slip and twinning deformation mechanisms should be undertaken.
- iii) Since final failure of Zr-2.5Nb pressure tube material occurs through ductile fracture, the present constitutive relations should be combined with a continuum-based damage model to predict ductile failure.

- iv) The Zerilli-Armstrong hcp constitutive relation with the parameters determined in this research should be applied to model pressure tube rupture experiments.

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Table 2.1

Chemical Specification For CANDU Zr-2.5Nb Pressure Tubes
Fabricated Prior to 1987

ELEMENT	SPECIFICATION
Niobium	2.4 - 2.8 wt. %
Oxygen	900 - 1300 ppm ²
Carbon	≤ 270 ppm
Chlorine	---
Chromium	≤ 200 ppm
Hydrogen	≤ 25 ppm ³
Iron	≤ 1500 ppm
Nickel	≤ 70 ppm
Nitrogen	≤ 65 ppm
Phosphorus	---
Silicon	≤ 120 ppm
Tantalum	≤ 200 ppm
Zirconium ¹	balance

¹ including other impurities

² ppm by weight

³ 20 ppm for ingot, 25 ppm for final tube

Table 2.2

Average Temperatures and Fluences for Tube P3J09

bundle # (from inlet)	axial position (m)	temperature (°C)	axial position (m)	fluence (x 10 ²⁵ n/m ²)
1	0.70	249.2	0.46	2.37
2	1.20	252.2	0.95	5.40
3	1.69	257.0	1.45	7.15
4	2.19	261.4	1.94	7.38
5	2.68	266.0	2.44	7.57
6	3.18	270.5	2.93	7.55
7	3.68	275.0	3.43	7.55
8	4.17	279.3	3.92	7.56
9	4.67	283.5	4.42	7.37
10	5.16	287.5	4.91	7.14
11	5.66	290.4	5.41	5.15
12	6.15	291.6	5.90	2.12

Table 2.3

Average Temperatures and Fluences for Tube B2N06

bundle # (from inlet)	axial position (m)	temperature (°C)	axial position (m)	fluence (x 10²⁵ n/m²)
1	0.43	250.8	0.18	0.20
2	0.92	253.7	0.67	2.21
3	1.42	258.0	1.17	4.02
4	1.91	262.2	1.66	5.42
5	2.41	267.8	2.16	6.55
6	2.90	274.0	2.66	7.46
7	3.40	279.9	3.15	8.32
8	3.89	285.1	3.65	8.31
9	4.39	289.4	4.14	7.50
10	4.88	292.3	4.64	6.10
11	5.38	293.8	5.13	4.50
12	5.87	294.0	5.63	2.50
13	6.37	294.0	6.12	0.22

Table 2.4

Average Temperatures and Fluences for Tube P4F17

bundle # (from inlet)	axial position (m)	temperature (°C)	axial position (m)	fluence (x 10²⁵ n/m²)
1	0.70	249.5	0.46	0.33
2	1.20	252.2	0.95	0.77
3	1.69	255.7	1.45	1.03
4	2.19	259.9	1.94	1.18
5	2.68	264.5	2.44	1.32
6	3.18	269.2	2.93	1.36
7	3.68	273.8	3.43	1.35
8	4.17	278.0	3.92	1.31
9	4.67	281.8	4.42	1.17
10	5.16	285.1	4.91	1.02
11	5.66	287.5	5.41	0.74
12	6.15	288.5	5.90	0.31

Table 2.5

Year Pressure Tubes were Removed and Effective Full Power Hours of Operation

tube	ingot #	year removed	inlet end	EFPH
P3J09	203181	1984	west	83670
B2N06	383847	1986	east	58080
P4F17	377936	1975	west	15090

Table 2.6

Fluence ($\times 10^{25}$ n/m²) for Pressure Tube Sections used to Manufacture Irradiated High Strain Rate Tensile Specimens

tube	position (cm)	position from inlet (cm)	fluence ($\times 10^{25}$ n/m²)
P3J09	12.7 - 17.7	12.7 - 17.7	.7 - .9
P3J09	64.2 - 72.8	64.2 - 72.8	3.5 - 4.0
P3J09	73.1 - 76.1	73.1 - 76.1	4.0 - 4.2
P3J09	165.3 - 168.9	165.3 - 168.9	7.2 - 7.3
P3J09	171.9 - 175.5	171.9 - 175.5	7.3
P3J09	202.4 - 212.4	202.4 - 212.4	7.4
P3J09	222.0 - 226.5	222.0 - 226.5	7.5
P3J09	258.9 - 263.0	258.9 - 263.0	7.5
B2N06	97.4 - 100.6 - P-E	148.6 - 151.8	4.9 - 5.0
B2N06	117.3 - 120.9 - P-E	168.5 - 172.1	5.5 - 5.6
B2N06	121.2 - 124.8 - P-E	172.4 - 176.0	5.6
B2N06	125.1 - 130.0 - P-E	176.3 - 181.5	5.7 - 5.8
B2N06	281.6 - 288.3 - P-W	265.2 - 271.9	7.5 - 7.6
B2N06	277.7 - 281.3 - P-W	272.2 - 275.8	7.6
P4F17	258.9 - 263.0	258.9 - 263.0	1.3

Table 3.1

Quasi-Static Tensile Tests on As-Received Zr-2.5Nb Pressure Tube Material in the Longitudinal Direction

Test Number	Temp. (°C)	Strain Rate (s ⁻¹)	σ_{uts} (MPa)	ϵ_f (%)
E51026B	20	10 ⁻⁴	723	27.9
E51031A	20	10 ⁻⁴	726	29.8
E51128E	20	10 ⁻³	737	32.8
E51128F	20	10 ⁻³	726	29.0
E51124A	20	10 ⁻²	751	29.2
E51124B	20	10 ⁻²	771	28.3
E51114A	20	10 ⁻¹	772	32.1
E51114B	20	10 ⁻¹	781	33.3
E51027A	250	10 ⁻⁴	545	30.6
E51027B	250	10 ⁻⁴	540	32.7
E51023A	300	10 ⁻⁴	532	30.4
E51024A	300	10 ⁻⁴	536	32.2
E60228A	300	10 ⁻³	570	no data
E60228B	300	10 ⁻³	565	no data
E60227B	300	10 ⁻²	568	no data
E60227C	300	10 ⁻²	560	no data
E60227D	300	10 ⁻²	579	no data
E51115B	300	10 ⁻¹	575	38.9

Table 3.2

Quasi-Static Tensile Tests on As-Received Zr-2.5Nb Pressure Tube Material in the Transverse Direction

Test Number	Temp. (°C)	Strain Rate (s ⁻¹)	σ_{uts} (MPa)	ϵ_f (%)
E51025B	20	.0001	792	26
E51026A	20	.0001	825	23.3
E51128C	20	.001	819	25.5
E51128D	20	.001	828	25.6
E51128A	20	.01	850	24
E51128B	20	.01	843	26.3
E51114C	20	.1	892	23.2
E51114D	20	.1	913	24.6
E51115A	20	.1	914	23.8
E51030A	250	.0001	624	21.2
E51030B	250	.0001	617	23.4
E51024B	300	.0001	612	20.6
E51025A	300	.0001	618	22.7
E60228C	300	.001	630	no data
E60228D	300	.001	624	no data
E60226B	300	.01	638	no data
E60227A	300	.01	644	no data
E51117A	300	.1	662	28.3
E51120A	300	.1	647	21.4

Table 3.3

High Strain Rate Tests Performed on As-Received
Zr-2.5Nb Pressure Tube Material in the Longitudinal Direction

Test Number	Temp. (°C)	Strain Rate (s ⁻¹)	σ_{us} (MPa)	ϵ_f (%)
E50504a	20	440	870	24.728
E50509c	20	460	900	23.026
E50425a	20	530	871	23.917
E50323c	20	615	883	22.991
E50323d	20	645	852	23.102
E50323b	20	680	844	24.55
E50503a	20	950	879	22.34
E50327a	20	1060	874	24.648
E50404a	20	1090	892	24.047
E50328b	20	1100	858	24.115
E50503b	20	1530	890	23.64
E50404b	20	1550	894	23.214
E50404c	20	1580	892	25.12
E50508b	20	1890	904	24.33
E50508c	20	2650	949	24.49
E50621C	71	500	835	21.5
E50620A	71	520	828	22.2
E50621E	71	940	846	20.8
E50621D	71	1025	859	21.5
E50621F	71	1455	876	20.6
E50621G	71	1585	839	18.8
E50621H	71	1930	852	18.8

...continued

Table 3.3 (concluded)

Test Number	Temp. (°C)	Strain Rate (s ⁻¹)	σ_{uts} (MPa)	ϵ_f (%)
E50607C	113	500	808	24.4
E50607E	113	515	813	22.2
E50607D	113	515	819	25.2
E50607B	113	560	821	22.1
E50608C	113	1000	828	21.6
E50608B	113	1020	838	22.1
E50608A	113	1045	817	21.6
E50608D	113	1490	854	22.6
E50609A	113	1815	843	23.7
E50609B	113	1860	848	23.6
E60223a	290	560	690	27.5
E60223b	290	590	672	27.8
E60223d	290	1090	700	27.0
E60223c	290	1120	692	27.7
E60308a	290	1550	701	26.9
E60311a	290	1700	720	25.5
E60312b	290	1860	715	25.4

Table 3.4

**High Strain Rate Tests on As-Received
Zr-2.5Nb Pressure Tube Material in the Transverse Direction**

Test Number	Temp. (°C)	Strain Rate (s ⁻¹)	σ_{uts} (MPa)	ϵ_f (%)
E50509b	20	400	1024	15.94
E50510a	20	400	1018	17.30
E50510b	20	415	1056	18.65
E50509a	20	460	1041	18.12
E50517a	20	540	1044	17.94
E50517b	20	600	1058	17.33
E50518b	20	600	1089	15.82
E50518a	20	620	1095	15.73
E50511c	20	1000	1121	16.57
E50511b	20	1015	1152	15.40
E50511a	20	1020	1101	15.13
E50515c	20	1470	1153	20.56
E50515a	20	1500	1179	17.08
E50512b	20	1500	1171	15.40
E50512a	20	1540	1173	17.28
E50516b	20	1640	1180	15.70
E50516a	20	1850	1170	18.41
E50515b	20	1890	1165	16.81
E50622A	71	450	1033	16.3
E50622B	71	1020	1051	15.9
E50612B	113	485	970	16.6
E50612C	113	488	995	15.1
E50612D	113	530	978	18.2
E50612A	113	626	1020	15.8
E50613B	113	1005	1028	15.4
E50613C	113	1017	1048	16.2
E50613A	113	1029	1068	16.0
E50613D	113	1417	1076	ND
E50614A	113	1475	1052	17.1

continued

Table 3.4 (concluded)

Test Number	Temp. (°C)	Strain Rate (s ⁻¹)	σ_{uts} (MPa)	ϵ_f (%)
E50613E	113	1479	1070	15.4
E50614B	113	1743	1124	14.2
E50614C	113	1822	1098	13.7
E50614E	113	1838	1084	15.6
E50614D	113	1838	1132	13.8
E60215a	290	390	762	19.2
E60220b	290	480	761	20.0
E60216a	290	620	766	18.9
E60219a	290	900	865	16.1
E60220a	290	960	841	18.6
E60219b	290	1080	872	19.1
E60216b	290	1130	828	19.3
E60219e	290	1530	879	17.9
E60219c	290	1930	915	19.2

Table 3.5

High Strain Rate Tensile Tests on Irradiated Zr-2.5Nb Pressure Tube Material in the Longitudinal Direction at Room Temperature

Specimen Name	Tube	Position from inlet (cm)	Fluence ($\times 10^{25} \text{ n/m}^2$)	Strain Rate (s^{-1})	σ_{uts} (MPa)	ϵ_f (%)
E70520A	P3J09	p-w 12.7 - 17.7	0.8	548	1102	20.0
E70521A	P3J09	p-w 12.7 - 17.7	0.8	574	1102	18.3
E70627A	P3J09	p-w 12.7 - 17.7	0.8	1314	1068	18.1
E70618A	P3J09	p-w 12.7 - 17.7	0.8	1360	1105	14.6
E70626A	P3J09	p-w 12.7 - 17.7	0.8	1753	1179	25
E80323A	P3J09	p-w 70.0 - 72.8	4.0	447	1055	13.7
E71209B	P3J09	p-w 64.2 - 72.8	3.6	524	1116	13.9
E80304A	P3J09	p-w 64.2 - 67.0	3.6	1955	1191	13.0
E80325A	B2N06	p-w 117.3 - 120.9	4.1	1973	1144	15.3
E80323D	P3J09	p-w 70.0 - 72.8	4.0	2309	1196	10.4
E70523D	P3J09	p-w 222.0 - 226.5	7.5	549	1063	19.0
E70523C	P3J09	p-w 222.0 - 226.5	7.5	615	1119	19.8
E70610A	P3J09	p-w 222.0 - 226.5	7.5	1384	1069	15.7
E70617B	P3J09	p-w 222.0 - 226.5	7.5	1872	1139	20.6
E80304B	B2N06	p-w 277.7 - 281.3	7.7	1965	1189	12.5
E80311C	P3J09	p-w 165.3 - 168.9	7.3	1999	1168	13.7

Table 3.6

High Strain Rate Tensile Tests on Irradiated Zr-2.5Nb Pressure Tube Material in the Longitudinal Direction at Elevated Temperatures

Specimen Name	Tube	Position from inlet (cm)	Fluence ($\times 10^{25} \text{ n/m}^2$)	temp ($^{\circ}\text{C}$)	Strain Rate (s^{-1})	σ_{uts} (MPa)	ϵ_f (%)
E80320B	P3J09	p-w 202.4 - 212.4	7.4	200	427	940	15.7
E80320C	P3J09	p-w 12.7 - 17.7	7.3	200	2021	1010	14.8
E80226B	B2N06	p-w 117.3 - 120.9	4.1	250	2112	950	14.5
E80226A	B2N06	p-w 117.3 - 120.9	4.1	250	2342	1063	17.5
E80212A	P3J09	p-w 12.7 - 16.3	.8	290	695	787	14.8
E80206A	P3J09	p-w 12.7 - 16.3	.8	290	747	784	21.5
E80225A	P3J09	p-w 12.7 - 16.3	.8	290	1042	816	21.4
E71212A	P3J09	p-w 73.1 - 76.7	4.1	290	433	808	13.7
E80306A	P3J09	p-w 73.1 - 76.7	4.1	290	2104	1002	13.8
E80217B	P3J09	p-w 73.1 - 76.7	4.1	290	2112	848	16.6
E80311A	B2N06	p-w 117.3 - 120.9	4.1	290	2125	967	14.8
E80217A	P3J09	p-w 73.1 - 76.7	4.1	290	2148	1010	17.0
E80310B	P3J09	p-w 165.3 - 168.9	7.2	290	451	856	19.8
E80305C	B2N06	p-w 277.7 - 281.3	7.7	290	457	817	15.9
E80323B	P3J09	p-w 165.3 - 168.9	7.2	290	469	874	13.0
E80306C	P3J09	p-w 165.3 - 168.9	7.2	290	476	833	18.0
E80310C	B2N06	p-w 277.7 - 281.3	7.7	290	2010	920	6.9

Table 3.7

High Strain Rate Tensile Tests on Irradiated Zr-2.5Nb Pressure Tube Material in the Transverse Direction at Room Temperature

Specimen Name	Tube	Position from inlet (cm)	Fluence ($\times 10^{25} \text{ n/m}^2$)	Strain Rate (s^{-1})	σ_{us} (MPa)	ϵ_f (%)
E70516A	P4F17	p-w 258.9 - 263.0	1.3	347	1450	15.2
E70522A	P3J09	p-w 12.7 - 17.7	0.8	437	1259	15.4
E70522B	P3J09	p-w 12.7 - 17.7	0.8	463	1205	13.0
E70515A	P4F17	p-w 258.9 - 263.0	1.3	578	1507	11.4
E70623B	P3J09	p-w 12.7 - 17.7	0.8	803	1233	12.6
E70618B	P3J09	p-w 12.7 - 17.7	0.8	819	1360	12.0
E70624B	P3J09	p-w 12.7 - 17.7	0.8	1641	1310	9.5
E70624C	P3J09	p-w 12.7 - 17.7	0.8	1674	1407	15.1
E80227A	B2N06	p-w 97.4 - 100.6	3.4	416	1182	17.1
E80219A	B2N06	p-w 97.4 - 100.6	3.4	469	1174	18.3
E80108A	B2N06	p-w 121.2 - 124.8	4.2	567	1077	12.7
E80213A	B2N06	p-w 121.2 - 124.8	4.2	586	1191	17.6
E80227B	B2N06	p-w 97.4 - 100.6	3.4	1914	1539	15.9
E80227C	B2N06	p-w 97.4 - 100.6	3.4	1935	1575	10.8
E80303E	P3J09	p-w 165.3 - 168.9	7.3	318	1254	13.9
E80303D	P3J09	p-w 165.3 - 168.9	7.3	360	1290	12.6
E70523A	P3J09	p-w 222.0 - 226.5	7.5	427	1273	14.1
E70523B	P3J09	p-w 222.0 - 226.5	7.5	439	1274	13.9
E70616A	P3J09	p-w 222.0 - 226.5	7.5	755	1255	13.1
E70616B	P3J09	p-w 222.0 - 226.5	7.5	935	1317	14.3
E70623A	P3J09	p-w 222.0 - 226.5	7.5	1581	1307	10.8
E70627B	P3J09	p-w 222.0 - 226.5	7.5	1634	1332	15.5
E80303B	P3J09	p-w 171.9 - 175.5	7.3	1752	1481	12.9
E71204B	B2N06	p-w 277.7 - 281.3	7.7	1872	1394	15.5

Table 3.8

High Strain Rate Tensile Tests on Irradiated Zr-2.5Nb Pressure Tube Material in the Transverse Direction at 290°C

Specimen Name	Tube	Position from inlet (cm)	Fluence ($\times 10^{25} \text{ n/m}^2$)	Strain Rate (s^{-1})	σ_{uts} (MPa)	ϵ_f (%)
E80309A	P3J09	p-w 12.7 - 16.3	0.8	505	946	16.2
E80309B	P3J09	p-w 12.7 - 16.3	0.8	2114	1037	10.3
E80219B	P3J09	p-w 73.1 - 76.7	4.1	705	982	12.1
E80219C	P3J09	p-w 73.1 - 76.7	4.1	823	970	10.4
E80324B	B2N06	p-w 97.4 - 100.6	3.4	1062	973	16.4
E80324A	B2N06	p-w 97.4 - 100.6	3.4	1145	1018	14.7
E80313A	B2N06	p-w 97.4 - 100.6	3.4	1781	1157	17.1
E80305B	P3J09	p-w 73.1 - 76.7	4.1	2034	1295	14.9
E80305A	P3J09	p-w 73.1 - 76.7	4.1	2080	1389	13.5
E80323E	B2N06	p-w 97.4 - 100.6	3.4	2107	1156	15.3
E80306D	P3J09	p-w 73.1 - 76.7	4.1	2165	1127	11.0
E80219D	P3J09	p-w 73.1 - 76.7	4.1	2330	1044	11.1
E80210B	B2N06	p-w 277.7 - 281.3	7.7	540	956	15.6
E80210A	B2N06	p-w 277.7 - 281.3	7.7	566	890	18.2
E80310A	P3J09	p-w 165.3 - 168.9	7.3	616	888	10.4
E80302A	P3J09	p-w 165.3 - 168.9	7.3	631	1000	15.5
E80302B	P3J09	p-w 165.3 - 168.9	7.3	651	933	14.8
E80112A	B2N06	p-w 277.7 - 281.3	7.7	705	936	18.4
E80304C	B2N06	p-w 277.7 - 281.3	7.7	1998	1129	12.3
E80303A	P3J09	p-w 171.9 - 175.5	7.3	2011	1137	14.0
E80303C	P3J09	p-w 165.3 - 168.9	7.3	2092	1107	19.2
E80309C	B2N06	p-w 277.7 - 281.3	7.7	2333	1100	15.7

Table 4.1

Johnson-Cook Parameters for the As-received Zr-2.5Nb Pressure Tube Material

Parameter	Isothermal Fit	Adiabatic Fit	Uncertainty
<i>A</i> (MPa)	585	420	34.7
<i>B</i> (MPa)	408	521	31.9
<i>n</i>	0.298	0.184	1.81×10^{-2}
<i>C</i>	1.00×10^{-2}	1.97×10^{-2}	1.35×10^{-4}
<i>m</i>	0.698	0.744	2.47×10^{-3}

Table 4.2

Johnson-Cook Parameters for the Irradiated Zr-2.5Nb Pressure Tube Material

Parameter	Adiabatic Fit	Uncertainty
<i>A</i> (MPa)	835	36.7
<i>B</i> (MPa)	691	60.9
<i>n</i>	0.472	0.103
<i>C</i>	1.10×10^{-2}	2.65×10^{-4}
<i>m</i>	0.728	5.40×10^{-3}

Table 4.3

Yield Stress and Strain Hardening Coefficients for As-received Zr-2.5Nb Pressure Tube
Material as a Function of Strain Rate and Temperature

strain rate (s ⁻¹)	temperature (K)	σ_y (MPa)	K (MPa)
0.0001	293	671	364
0.001	293	649	397
0.01	293	695	371
0.1	293	694	454
562	293	833	267
1050	293	868	163
1553	293	907	139
1890	293	878	212
2650	293	922	116
510	344	796	223
983	344	865	80
1520	344	896	17
1930	344	880	8
523	386	799	175
1022	386	837	97
1490	386	857	128
1838	386	873	54
560	563	657	213
590	563	633	227
1090	563	695	134
1120	563	687	144
1550	563	698	76
1700	563	757	8
1860	563	733	6
0.0001	573	499	228
0.1	573	485	387

Table 4.4

Zerilli-Armstrong Parameters for the As-received Zr-2.5Nb Pressure Tube Material

Parameter	Isothermal Fit	Adiabatic Fit	Uncertainty
σ_a (MPa)	386	443	6.03
B (MPa)	897	787	13.3
β_0 (K ⁻¹)	2.90×10^{-3}	3.46×10^{-3}	6.71×10^{-5}
β_1 (K ⁻¹)	1.40×10^{-4}	1.59×10^{-4}	3.53×10^{-6}
B_0 (MPa)	570	766	25.4
α_0 (K ⁻¹)	8.30×10^{-4}	1.84×10^{-3}	8.64×10^{-5}
α_1 (K ⁻¹)	6.01×10^{-5}	2.01×10^{-5}	3.50×10^{-6}

Table 4.5

Zerilli-Armstrong Parameters for the Irradiated Zr-2.5Nb Pressure Tube Material

Parameter	Adiabatic Fit	Uncertainty
σ_a (MPa)	644	134
B (MPa)	628	266
β_0 (K ⁻¹)	3.60×10^{-3}	3.64×10^{-3}
β_1 (K ⁻¹)	0	1.06×10^{-5}
B_0 (MPa)	3242	537
α_0 (K ⁻¹)	4.70×10^{-3}	3.91×10^{-4}
α_1 (K ⁻¹)	2.06×10^{-4}	1.55×10^{-5}

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Figure 1.1: Schematic of a CANDU Reactor Assembly.

Figure 1.2: Schematic of a CANDU Fuel Channel.

Double-Arc-Melted Ingot
Ultrasonic Inspection
Forge to Bar
Machine Hollow Billets
Ultrasonic Inspection
Clad Billets in Steel and Copper
Preheat Billets to 850°C and Extrude into Tube
Dissolve the Cladding
Sand Blast Inside Surface and Grind Outside
Cold-Draw to ~ 15%
Sand Blast Inside Surface and Grind Outside
Cold-Draw to 10 to 15%
Hydrostatic Testing
Hone Inside Surface and Grind Outside
Ultrasonic Inspection
Stress Relieve in Autoclave for 24 h at 400 °C
Straighten and Correct Ovality
Ultrasonic Inspection

Figure 1.3: The Fabrication Process of Zr-2.5Nb Pressure Tubes.

Figure 1.4: Allotropic Transformation From Hexagonal-Close-Packed α - Zirconium To Body-Centered-Cubic β -Zirconium.

Figure 1.5: Principal Crystallographic Planes For α -Zirconium.

a)

b)

Figure 1.6: Micrographs of As-Received Zr-2.5Nb Pressure Tube Material (X 1000 Mag.), a) Longitudinal-Radial Plane, b) Transverse-Radial Plane.

Figure 1.7: Pole Figures For the (0001) and $(1\bar{1}00)$ Planes of As-Received Zr-2.5Nb Pressure Tube Material.

Figure 1.8: Schematic of The Idealized Texture of As-Received Zr-2.5Nb Pressure Tube Material.

Figure 1.9: The Slip Systems in α -Zirconium.

Figure 1.10: The Twin Systems in α -Zirconium.

Figure 1.11: The Relationship Between The Direction of Applied Force and The Shearing Stress Acting on a Deformation System for a Single-Crystal Rod.

Figure 1.12: The Schmid Factor Orientation Dependence for Various Slip and Twin Systems in Zirconium (According To Reference 21).

Figure 1.13: Schematic Representation of a Displacement Spike [64].

Figure 1.14: Schematic Representation of a Seeger Zone [65].

Figure 1.15: Schematic Representation of a) Interstitial and b) Vacancy Dislocation Loops.

Figure 1.16: Ultimate Tensile Strength of Zr-5Nb Pressure Tubes With and Without Irradiation (fluence was $1.27 \times 10^{25} \text{ n/m}^2$ at $\sim 300 \text{ C}$) [81].

Figure 1.17: Ultimate Tensile Strength at $\sim 300 \text{ C}$ of Zr-2.5Nb Pressure Tubes as a Function of Fluence [81].

Figure 1.18: Ductility of Zr-2.5Nb Pressure Tubes With and Without Irradiation, (fluence was $1.27 \times 10^{25} \text{ n/m}^2$ at $\sim 300 \text{ C}$) [81].

Figure 1.19: Reduction in Area at Failure of Zr-2.5Nb Pressure Tubes as a Function of Fluence, at $\sim 300 \text{ C}$ [81].

Figure 1.20: Typical Stress-Strain Responses for Irradiated Longitudinal and Transverse Specimens Tested at 250 C in [74].

Figure 1.21: Effect of Temperature on the Yield Stress and Uniform Elongation for Longitudinal and Transverse Specimens Tested in [74].

Figure 1.22: Effect of Strain Rate and Temperature on UTS of Zr-2.5Nb Pressure Tube Material [81, 84].

Figure 1.23: Schematic Representation of the Relationship Between the Flow Stress and the Logarithm of the Strain Rate [88].

Figure 1.24: Bilinear Variation of Flow Stress at 5% Plastic Strain with Logarithm of Strain Rate [97].

Figure 1.25: Schematic Representation of a Split Hopkinson Bar.

Figure 1.26: Schematic Representation of a Split Hopkinson Bar Specimen Attached to the Ends of the Input and Output Bars.

(a)

(b)

Figure 1.27: Details of a Tensile Split Hopkinson Bar Specimen From [1].

Figure 1.28: Schematic of Four Split Hopkinson Bar Configurations Utilizing Tension-Loading Techniques.

Figure 1.29: Details of a Tensile Split Hopkinson Bar Specimen From [4].

Figure 2.1: Schematic of the Direct-Tension Split Hopkinson Bar.

Figure 2.2: Photograph of the Direct-Tension Split Hopkinson Bar.

Figure 2.3: Photographs of the Clamp/Release Mechanism on the Direct-Tension Split Hopkinson Bar.

Figure 2.4: Photograph of the Loading Mechanism on the Direct-Tension Split Hopkinson Bar.

Figure 2.5: Schematic of the Tensile Specimen Design.

Figure 2.6: Photograph of the Tensile Specimen Design.

Figure 2.7: Photograph of the Tensile Specimen Attached to the Adapter.

Figure 2.8: Photograph of the Tensile Specimen Attached to the Split Hopkinson Bar.

Figure 2.9: Schematic of the Heating Chamber.

Figure 2.10: Photograph of the Heating Chamber.

Figure 2.11: Reduction in Cross-Sectional Area as a Function of Engineering Strain for As-Received Material.

Figure 2.12: Reduction in Cross-Sectional Area as a Function of Engineering Strain Irradiated Material.

Figure 3.1: Typical Engineering Stress-Strain Responses of As-received Longitudinal Specimens Tested at Strain Rates of $1 \times 10^{-4} \text{ s}^{-1}$ and 0.1 s^{-1} , and at Room Temperature (RT) and 300°C .

Figure 3.2: Typical Engineering Stress-Strain Responses of As-received Transverse Specimens Tested at Strain Rates of $1 \times 10^{-4} \text{ s}^{-1}$ and 0.1 s^{-1} , and at Room Temperature (RT) and 300°C .

Figure 3.3: Comparison of Engineering Strain Rate as a Function of Engineering Strain.

Figure 3.4: Comparison of the Engineering Stress-Strain Responses of As-received Longitudinal Specimens Tested at Room Temperature (RT) and 290°C.

Figure 3.5: The UTS as a Function of Log-Strain Rate for the As-received Longitudinal Specimens.

Figure 3.6: Effect of Strain Rate on Failure Strains for As-received Longitudinal Specimens.

Figure 3.7: Comparison of the Engineering Stress-Strain Responses of As-received Transverse Specimens Tested at Room Temperature (RT) and 290°C.

Figure 3.8: The UTS as a Function of Log-Strain Rate for the As-received Transverse Specimens.

Figure 3.9: Effect of Strain Rate on Failure Strains for As-received Transverse Specimens.

Figure 3.10: Typical High Strain Rate Engineering Stress-Strain Responses of Irradiated Longitudinal Specimens Tested at Room Temperature (RT) and 290°C.

Figure 3.11: Comparison of Engineering Stress Strain Responses of As-received and Irradiated Longitudinal Specimens Tested at Room Temperature.

Figure 3.12: Comparison of Engineering Stress Strain Responses of As-received and Irradiated Longitudinal Specimens Tested at 290°C.

Figure 3.13: The UTS as a Function of Log-Strain Rate for As-received and Irradiated Longitudinal Specimens at Room Temperature.

Figure 3.14: The UTS as a Function of Log-Strain Rate for As-received and Irradiated Longitudinal Specimens at 290 to 300°C.

Figure 3.15: The UTS as a Function of Log-Strain Rate for Irradiated Longitudinal Specimens at Room Temperature and 290°C.

Figure 3.16: Failure Strain as a Function of Strain Rate for As-received and Irradiated Longitudinal Specimens at Room Temperature.

Figure 3.17: Failure Strain as a Function of Strain Rate for As-received and Irradiated Longitudinal Specimens at 290°C.

Figure 3.18: Typical High Strain Rate Engineering Stress-Strain Responses of Irradiated Transverse Specimens Tested at Room Temperature (RT) and 290°C.

Figure 3.19: Comparison of Engineering Stress-Strain Responses of As-received and Irradiated Transverse Specimens Tested at Room Temperature.

Figure 3.20: Comparison of Engineering Stress-Strain Responses of As-received and Irradiated Transverse Specimens Tested at 290°C.

Figure 3.21: The UTS as a Function of Log-Strain Rate for As-received and Irradiated Transverse Specimens at Room Temperature.

Figure 3.22: The UTS as a Function of Log-Strain Rate for As-received and Irradiated Transverse Specimens at 290 C.

Figure 3.23: The UTS as a Function of Log-Strain Rate for Irradiated Transverse Specimens at Room Temperature (RT) and 290°C.

Figure 3.24: Failure Strain as a Function of Strain Rate for As-received and Irradiated Transverse Specimens at Room Temperature.

Figure 3.25: Failure Strain as a Function of Strain Rate for As-received and Irradiated Transverse Specimens at 290°C.

Figure 3.26: Effect of Strain Rate on Strength of Irradiated Zr-2.5Nb Pressure Tube Material in Longitudinal and Transverse Directions at Room Temperature.

Figure 3.27: Effect of Strain Rate on Strength of Irradiated Zr-2.5Nb Pressure Tube Material in Longitudinal and Transverse Directions at 290 to 300°C.

Figure 3.28: Stress-Strain Responses of Longitudinal As-Received Specimens Tested at Carleton University.

Figure 3.29: Stress-Strain Response of Transverse As-Received Specimen Tested at Carleton University.

(a)

(b)

Figure 3.30: Photographs of specimen E51026B (a) longitudinal-radial plane
(b) longitudinal-transverse plane (RT, 10^{-4} s^{-1}) (x 30 mag.)

(a)

(b)

Figure 3.31: Photographs of specimen E51024A (a) longitudinal-radial plane (b) longitudinal-transverse plane ($290\text{ }^{\circ}\text{C}$, 10^{-4} s^{-1}) (x 30 mag.)

(a)

(b)

Figure 3.32: Photographs of specimen E50508C (a) longitudinal-radial plane
(b) longitudinal-transverse plane (RT, 2650 s^{-1}) (x 30 mag.)

(a)

(b)

Figure 3.33: Photographs of specimen E60311A (a) longitudinal-radial plane
(b) longitudinal-transverse plane (290°C , 1700 s^{-1}) (x 30 mag.)

(a)

(b)

Figure 3.34: Photographs of specimen E50511C (a) transverse-radial plane
(b) transverse-longitudinal plane (RT, 1000 s^{-1}) (x 30 mag.)

(a)

(b)

Figure 3.35: Photographs of specimen C71217A (a) longitudinal-radial plane
(b) longitudinal-transverse plane (RT, 1700 s^{-1}) (x 30 mag.)

(a)

(b)

Figure 3.36: Photographs of irradiated specimen E80323D (a) longitudinal-radial plane (b) longitudinal-transverse plane ($F \sim 4.1 \times 10^{25} \text{ n/m}^2$, RT, 2310 s^{-1}) (x 30 mag.)

(a)

(b)

Figure 3.37: Photographs of irradiated specimen E80217B (a) longitudinal-radial plane (b) longitudinal-transverse plane ($F \sim 4.1 \times 10^{25}$, 290°C , 2110 s^{-1}) (x 30 mag.)

Figure 4.1: True Stress as a Function of Plastic Strain for Experiment E50327A.

Figure 4.2: Strain Rate as a Function of Plastic Strain for Experiment E50327A.

Figure 4.3: Temperature as a Function of Plastic Strain for Experiment E50327A.

Figure 4.4: Stress-Strain Responses as a Function of Strain Rate and temperature for As-Received Zr-2.5Nb Material.

Figure 4.5: Flow Stress as a Function of Log-Strain Rate As-Received Zr-2.5Nb Material.

Figure 4.6: Stress-Strain Responses as a Function of Strain Rate and temperature for Irradiated Zr-2.5Nb Material.

Figure 4.7: Flow Stress as a Function of Log-Strain Rate Irradiated Zr-2.5Nb Material.

Figure 4.8: Predicted Stress-Strain Curves from Johnson-Cook Isothermal Fit Compared to Experimental Data for As-Received Material.

Figure 4.9: Predicted Stress-Strain Curves from Johnson-Cook Adiabatic Fit Compared to Experimental Data for As-Received Material.

Figure 4.10: Linear Fits of True Stress as a Function of Square Root-Plastic Strain for As-Received Material.

Figure 4.11: Yield Stress as a Function of ξ_β for As-received Material.

Figure 4.12: Strain Hardening Coefficient as a Function of ξ_α for As-Received Material.

Figure 4.13: Predicted Stress-Strain Curves from Zerilli-Armstrong Isothermal Fit Compared to Experimental Data for As-Received Material.

Figure 4.14: Predicted Stress-Strain Curves from Zerilli-Armstrong Adiabatic Fit Compared to Experimental Data for As-Received Material.

Figure 4.15: Predicted Stress-Strain Curves from Johnson-Cook Adiabatic Fit Compared to Experimental Data for Irradiated Material.

Figure 4.16: Predicted Stress-Strain Curves from Zerilli-Armstrong Adiabatic Fit Compared to Experimental Data for Irradiated Material.

Figure 5.1: Comparison of Analytical and Single-Element Calculations for the Johnson-Cook and Zerilli-Armstrong hcp Constitutive Relations.

Figure 5.2: Finite Element Mesh of the Direct-Tension Split Hopkinson Specimens.

Figure 5.3: Gauge Length Region of the Finite Element Mesh.

Figure 5.4: Example of a Velocity-Time History Used for Velocity Boundary Condition in Finite Element Calculations.

Figure 5.5: Comparison of the Measured and Predicted Engineering Stress-strain Curves for As-Received Zr-2.5Nb Pressure Tube Material.

Figure 5.6: Comparison of the Measured and Predicted Engineering Stress-strain Curves for Irradiated Zr-2.5Nb Pressure Tube Material.

Figure 5.7: Localized Deformation in a) transverse and b) radial directions at 19% Engineering Strain for Simulation of Room Temperature Test on As-Received Material at Strain Rate of 1580 s^{-1} .

Figure 5.8: Localized Deformation in a) transverse and b) radial directions at 25% Engineering Strain for Simulation of Room Temperature Test on As-Received Material at Strain Rate of 1580 s^{-1} .

Figure 5.9: Comparison of the Predicted and Measured Reduction in Cross-Sectional Area for As-Received Zr-2.5Nb Pressure Tube Material.

a

b

Figure 5.10: Comparison of the Predicted and Actual Deformation at 19% Engineering Strain for As-Received Zr-2.5Nb Pressure Tube Material Tested at Room Temperature and $\sim 1700 \text{ s}^{-1}$ (longitudinal specimen).

Appendix A

Split Hopkinson Bar Analysis Program (SHB_ANALYSIS)

```

PROGRAM SHB_ANALYSIS
C
C *****
C   Program is used to analyze tensile SHB test results
C *****
C
C                               version 1.1
C
C *****
C
REAL*8 TR(10001),SR(10001),ST(10001),DELTAT
REAL*8 RATE(10001),STRAIN(10001),STRESS(10001)
REAL*8 TSTRAIN(10001),ESTRAIN(10001),PSTRAIN(10001)
REAL*8 TSTRESS(10001),TEMP(10001),TRATE(10001),TAREAS(10001)
REAL*8 ZEROIR,ZEROXR,ZEROYR,HOMOTEMP
REAL*8 ZEROIT,ZEROXT,ZEROYT
REAL*8 CONX,CONYR,CONYT,F,E,CONS,CONT
REAL*8 LENGTH,WAVE,DIAM,PI,AREAB,EBAR,AREAS
REAL*8 RA1,RA2,RA3,REDAREA,EMOD(10001)
REAL*8 SUM,WIDTH,THICK,TEMPO
REAL*8 MXSTRAIN,YSTRAIN,ZSTRAIN
REAL*8 ZRDENS,SPECHEAT,ROOTPS

CHARACTER*32 FILEIN,FILEOUT,DATFILE,LOGFILE
CHARACTER*1 ANSWER

INTEGER NDATA,LASTN,N,I,J,STEP,ORIENT

C *****

ZRDENSITY = 6.53E+3
PI = 3.141593

C *****

WRITE(*,2)
2  FORMAT(/' SPLIT HOPKINSON BAR ANALYSIS PROGRAM SHB_ANAL ')
WRITE(*,3)
3  FORMAT(/' for Zr-2.5Nb Pressure Tube Specimens')
WRITE(*,4)
4  FORMAT(/'   version 1.1  98/02/17 ')

C ***** read in filename for input data file *****

5  WRITE(*,10)
10  FORMAT(/,'enter filename for log file (ex. EX1231A.LOG)')

```

```

& '$)
  READ(*,1050) LOGFILE

  OPEN(UNIT=3,FILE=LOGFILE,ERR=20,STATUS='NEW')
  GOTO 49

20 PRINT*,LOG file already exists!! try again '
  GOTO 5

C ***** read in filename for log file *****

49 WRITE(*,50)
50 FORMAT(/,'enter data file for experiment (ex. EX1231A.IN)
  & '$)
  READ(*,1050) DATFILE

  OPEN(UNIT=2,FILE=DATFILE,ERR=60,STATUS='OLD')
  GOTO 99

60 PRINT*,INPUT file does NOT exist!! try again '
  GOTO 49

C ***** start of data input from DAT files *****

99 NDATA=0

C ***** write program info. into LOGFILE *****

  WRITE(3,2)
  WRITE(3,3)
  WRITE(3,4)

  WRITE(3,*) 'input file DATFILE is ',DATFILE

C ***** read in if axial or circumferential specimen *****

  READ(2,*) ORIENT

  IF(ORIENT.EQ.0)THEN

    PRINT*,THIS IS A AXIAL SPECIMEN'
    WRITE(3,*)'specimen orientation is axial'
    GOTO 100

  ELSEIF(ORIENT.EQ.1)THEN

```

```

PRINT*, 'THIS IS A CIRCUMFERENTIAL SPECIMEN'
WRITE(3,*) 'specimen orientation is circumferential'
GOTO 100

ELSE

PRINT*, 'ERROR IN ', DATFILE
PRINT*, 'the specimen orientation must be defined !!'
STOP

ENDIF

C ***** enter zero time for reflected wave *****

100 READ(2,1000) ZEROXR
WRITE(3,*) 'zero time for the reflected wave ZEROXR is ', ZEROXR

C ***** read in reflected wave data *****

READ(2,1050) FILEIN

OPEN(UNIT=1, FILE=FILEIN, ERR=215, STATUS='OLD')
GOTO 220

215 PRINT*, 'cannot find reflected data file!'

220 WRITE(3,*) 'reflected data file FILEIN I is ', FILEIN

WRITE(*,221)
221 FORMAT('reading reflected strain data')

READ(1,*)
READ(1,*)

DO 225 I=1,9000
READ(1,*,END=230) CONS,CONT

IF (CONT.GE.ZEROXR) THEN

NDATA=NDATA+1

SR(NDATA) = CONS
TR(NDATA) = CONT
ENDIF

225 CONTINUE

```

```

230 CLOSE(1)

C ***** set zero strain for reflected wave data *****

      ZEROYR = SR(1)

C ***** enter zero time for transmitted wave *****

      READ(2,1000) ZEROXT
      WRITE(3,*)'zero time for the transmitted wave ZEROXT is ',ZEROXT

C ***** read in transmitted wave data *****

      READ(2,1050) FILEIN

      OPEN(UNIT=1,FILE=FILEIN,ERR=315,STATUS='OLD')
      GOTO 320

315 PRINT*,'cannot find transmitted data file!'
      STOP

320 WRITE(3,*)'transmitted data file FILEIN II is',FILEIN

      WRITE(*,321)
321 FORMAT('reading transmitted strain data')

      READ(1,*)
      READ(1,*)

      NDATA = 0

      DO 325 I=1,9000

      READ(1,*,END=330) CONS,CONT

      IF (CONT.GE.ZEROXT) THEN

          NDATA=NDATA+1

          ST(NDATA) = CONS
      ENDIF

325 CONTINUE

330 CLOSE(1)

```


C ***** set zero strain for transmitted wave data *****

ZEROYT = ST(1)

C ***** calculate the time step between data points *****

DELTAT = ABS(TR(2)-TR(1))

WRITE(3,*)'time between data points DELTAT is ',DELTAT

C ***** enter conversion data for time (us to s) *****

C ***** enter conversion data for strain (mV to m/m) *****

READ(2,1000) CONX

WRITE(3,*)'convert factor for time CONX is ',CONX

READ(2,1010) CONYR,CONYT

WRITE(3,*)'convert factor for reflected strain CONYR is ',CONYR

WRITE(3,*)'convert factor for transmitted strain CONYT is ',CONYT

C ***** enter conversion data for strain *****

C ***** gage factor, supply voltage (V)

READ(2,*) F,E

WRITE(3,*)'strain gauge factor F is ',F

WRITE(3,*)'strain gauge supply voltage E is ',E

C ***** convert time to us and strain (mV) to strain (m/m) *****

DO 635 I=1,NDATA

TR(I) = TR(I)*CONX

SR(I) = 2.*1.E+03*(SR(I)-ZEROYR)/((E*F/1.E+03)-(1.E+03*
& (SR(I)-ZEROYR)*F/1.E+06))

SR(I) = SR(I)*CONYR

ST(I) = 2.*1.E+03*(ST(I)-ZEROYT)/((E*F/1.E+03)-(1.E+03*
& (ST(I)-ZEROYT)*F/1.E+06))

ST(I) = ST(I)*CONYT

635 CONTINUE

C ***** enter data for bar *****

READ(2,1010) AREAB,EBAR

WRITE(3,*)'x-sectional area of shb AREAB is ',AREAB

```

WRITE(3,*)'elastic modulus of shb EBAR is ',EBAR

C ***** read file name for bar strains output *****

      READ(2,1050) FILEOUT
      WRITE(3,*)'output file containing shb strains FILEOUT I is ',FILEO
&UT

      OPEN(UNIT=1,FILE=FILEOUT,STATUS='UNKNOWN')

C ***** read step to skip output of data points *****

      READ(2,*) STEP
      WRITE(3,*)'skipped points on output STEP is ',STEP

      WRITE(1,642)

      DO 640 I=1,NDATA-1,STEP

      CONYR = SR(I)*EBAR
      CONYT = ST(I)*EBAR

      WRITE(1,645) TR(I),CONYR,CONYT

640 CONTINUE

      CLOSE(1)

642 FORMAT(2X,' time',4X,' reflected ',' transmitted')
645 FORMAT(2X,3E12.4)

C ***** enter data for specimen *****

      READ(2,1020) LENGTH,WIDTH,THICK
      WRITE(3,*)'length of shb specimen LENGTH is ',LENGTH
      WRITE(3,*)'width of shb specimen WIDTH is ',WIDTH
      WRITE(3,*)'thickness of shb specimen THICK is ', THICK

C ***** calculation of specimen cross-sectional area *****

      AREAS = THICK*WIDTH

C ***** calculation of Young's Modulus (C.K. Chow MMB-87-045) *****
C ***** NB: based on specimen orientation and temperature *****

      READ(2,1000) TEMP0

```

```

WRITE(3,*)'test temperature TEMP0 is ',TEMP0

IF(ORIENT.EQ.0)THEN

    EMOD(1) = 96.5633 - 1.40282E-05*TEMP0 - 2.91282E-04*(TEMP0**2.)
&    + 3.87923E-07*(TEMP0**3.)

    EMOD(1) = EMOD(1)*1.0E+09
    WRITE(3,*)'initial specimen elastic modulus EMOD(1) is ',EMOD(1)

ELSE

    EMOD(1) = 102.466 - 0.0117428*TEMP0 - 8.06869E-05*(TEMP0**2.)

    EMOD(1) = EMOD(1)*1.0E+09
    WRITE(3,*)'initial specimen elastic modulus EMOD(1) is ',EMOD(1)

ENDIF

C ***** calculate the elastic wave velocity *****

    READ(2,1000) WAVE
    WRITE(3,*)'wave speed in bar WAVE is ',WAVE

C ***** calculation of stress, strain, strain-rate *****
C ***** true stress and strain *****
C
C ***** read coefficients for area function 3-RD order poly. *****

    READ(2,1020) RA1,RA2,RA3
    WRITE(3,*)'1st poly. coef. for area RA1 is ',RA1
    WRITE(3,*)'2nd poly. coef. for area RA2 is ',RA2
    WRITE(3,*)'3rd poly. coef. for area RA3 is ',RA3

C ***** read factor for calculation of temp. increase due to plastic work *****

    READ(2,1000) TFACTOR
    WRITE(3,*)'% plastic work convert to heat TFACTOR is ',TFACTOR

    READ(2,1000) MXSTRAIN
    WRITE(3,*)'maximum strain until calculations stopped ',MXSTRAIN

C *****
C ***** start main loop for calculations *****
C *****

```

```

LASTN = 0

DO 700 I=1,NDATA

C      ***** calculation of engineering stress *****

      STRESS(I) = EBAR*AREAB*ST(I)/AREAS

C      ***** calculation of engineering strain rate *****

      RATE(I) = 2.*WAVE*SR(I)/LENGTH

C      ***** calculation of engineering stress *****

      IF (I.EQ.1) THEN
        STRAIN(I) = 0.
      ELSE
        SUM = ((RATE(I)+RATE(I-1))/2.)*
&          (TR(I)-TR(I-1))*1.0E-06
        STRAIN(I) = STRAIN(I-1)+SUM
      ENDIF

      IF(STRAIN(I).LE.MXSTRAIN) THEN

        LASTN = LASTN + 1

      ELSE

        GOTO710

      ENDIF

      REDAREA = RA1*STRAIN(I)+RA2*STRAIN(I)**2.+RA3*STRAIN(I)**3

      REDAREA = REDAREA/100.

      TAREAS(I) = AREAS*(1.-REDAREA)

C      ***** calculation of true stress *****

      TSTRESS(I) = STRESS(I)*(AREAS/TAREAS(I))

C      ***** calculation of true strain *****

      TSTRAIN(I) = abs(LOG(TAREAS(I)/AREAS))

```

```

C      ***** calculation of true strain rate *****

      IF (I.EQ.1) THEN

          TRATE(I) = 0.

      ELSE

          TRATE(I) = (TSTRAIN(I)-TSTRAIN(I-1))/DELTAT

      ENDIF

C ***** calc. of elastic and plastic strain *****

      ESTRAIN(I) = TSTRESS(I)/EMOD(I)

      PSTRAIN(I) = TSTRAIN(I)-ESTRAIN(I)

      IF (PSTRAIN(I).LT.0.) PSTRAIN(I) = 0.

C      ***** calculation of temperature *****

      IF (I.EQ.1) THEN

          TEMP(I) = TEMP0

      ELSE

          SPECHEAT = 1.13714E-04*TEMP(I-1)+0.2867
          SPECHEAT = SPECHEAT*1.0E+03

          SUM = ((TSTRESS(I)+TSTRESS(I-1))/2.)*
&      (PSTRAIN(I)-PSTRAIN(I-1))
          SUM = (TFACTOR*SUM)/(6.53E+03*SPECHEAT)

          TEMP(I) = TEMP(I-1)+SUM

      ENDIF

      IF(ORIENT.EQ.0)THEN

          EMOD(I+1) = 96.5633 - 1.40282E-05*TEMP(I) - 2.91282E-04*
&      (TEMP(I)**2.) + 3.87923E-07*(TEMP(I)**3.)

          EMOD(I+1) = EMOD(I+1)*1.0E+09

```

```

ELSE

EMOD(I+1) = 102.466-0.0117428*TEMP(I)-8.06869E-05*(TEMP(I)**2.)

EMOD(I+1) = EMOD(I+1)*1.0E+09

ENDIF

700 CONTINUE

C ***** enter output file for engineering data *****

710 READ(2,1050) FILEOUT
    WRITE(3,*)'output file for engineering data FILEOUT II is ',
    & FILEOUT

    OPEN(UNIT=20,FILE=FILEOUT,STATUS='UNKNOWN')

    WRITE(20,735)

    DO 730 I=1,LASTN

        STRESS(I) = STRESS(I)*1E-06

        WRITE(20,740) TR(I),RATE(I),STRAIN(I),STRESS(I),TAREAS(I),TEMP(I)
730 CONTINUE

735 FORMAT(2X,' time      strain-rate      strain      stress
    & area      temp')
740 FORMAT(2X,6E14.6)

    CLOSE(20)

C ***** enter output file for true stress/true strain data *****

    READ(2,1050) FILEOUT
    WRITE(3,*)'output file for strain rate vs. strain FILEOUT III is '
    &,FILEOUT

    OPEN(UNIT=20,FILE=FILEOUT,STATUS='UNKNOWN')

    WRITE(20,765)

    DO 760 I=1,LASTN

```

```

TSTRESS(I) = TSTRESS(I)/1.E+06

WRITE(20,770) TSTRAIN(I),TSTRESS(I)
760 CONTINUE

CLOSE(20)

765 FORMAT(2X,' t-strain    t-stress ')

770 FORMAT(2X,2E14.6)

C ***** output of data for Z-A constitutive modeling *****

READ(2,1050) FILEOUT
WRITE(3,*)'output file for Z-A data is FILEOUT IV
&is 'FILEOUT

OPEN(UNIT=20,FILE=FILEOUT,STATUS='UNKNOWN')

WRITE(20,795)
DO 790 I=1,LASTN

ROOTPS = PSTRAIN(I)**0.5
TEMPK = TEMP(I)+273

WRITE(20,1040) ROOTPS,TSTRESS(I),PSTRAIN(I),RATE(I),TEMPK
790 CONTINUE

CLOSE(20)

795 FORMAT(2X,' rootps    tstress    pstrain    rate
& temp ')

C ***** output of data for J-C constitutive modeling *****

READ(2,1050) FILEOUT
WRITE(3,*)'output file for true stress vs. true strain FILEOUT V i
&s 'FILEOUT

OPEN(UNIT=20,FILE=FILEOUT,STATUS='UNKNOWN')

WRITE(20,810)
DO 800 I=1,LASTN

C ***** calculate homologous temperature for constitutive modeling *****

```

```

        HOMOTEMP = (TEMP(I)+273)/(1760+273)

        WRITE(20,1040) PSTRAIN(I),TSTRESS(I),TRATE(I),HOMOTEMP
800  CONTINUE

        CLOSE(20)
810  FORMAT(2X,' P-strain   T-stress   T-rate   homotemp')
C
C
1000 FORMAT(2X,1E12.6)
1010 FORMAT(2X,2E14.6)
1020 FORMAT(2X,3E12.6)
1025 FORMAT(2X,3E14.6)
1030 FORMAT(2X,4E14.6)
1040 FORMAT(2X,5E14.6)
1050 FORMAT(A32)

        STOP
        END

```


Appendix B

Engineering Stress-Strain Responses for Longitudinal As-Received Zr-2.5Nb Pressure Tube Specimen at Quasi-Static Strain Rates

Appendix B1:
(Room Temperature Data)

Appendix B2:

(250 °C Data)

Appendix B3:

(300 °C Data)

Appendix C

Engineering Stress-Strain Responses for Transverse As-Received Zr-2.5Nb Pressure Tube Specimen at Quasi-Static Strain Rates

Appendix C1:
(Room Temperature Data)

Appendix C2:

(250 °C Data)

Appendix C3:

(300 °C Data)

Appendix D

Engineering Stress-Strain Responses for Longitudinal As-Received Zr-2.5Nb Pressure Tube Specimen at High Strain Rates

Appendix D1:
(Room Temperature Data)

Appendix D2:

(71 °C Data)

Appendix D3: (113 °C Data)

Appendix D4:

(290 °C Data)

Appendix E

Engineering Stress-Strain Responses for Transverse As-Received Zr-2.5Nb Pressure Tube Specimen at High Strain Rates

Appendix E1:
(Room Temperature Data)

Appendix E2:

(71 °C Data)

Appendix E3:

(113 °C Data)

Appendix E4:

(290 °C Data)

Appendix F

Engineering Stress-Strain Responses for Longitudinal Irradiated Zr-2.5Nb Pressure Tube Specimen at High Strain Rates

Appendix F1:
(Room Temperature Data)

Appendix F2:

(200 °C Data)

Appendix F3:

(250 °C Data)

Appendix F4:

(290 °C Data)

Appendix G

Engineering Stress-Strain Responses for Transverse Irradiated Zr-2.5Nb Pressure Tube Specimen at High Strain Rates

Appendix G1:
(Room Temperature Data)

Appendix G2:

(290 °C Data)