

Imperial College
London

A study of inelastic/anelastic behaviour
and springback of AA7050 alloy in creep
age forming

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This dissertation is submitted for the degree of Doctor of Philosophy

June 2017

Gloria in Excelsis Deo!

Declarations

This thesis is submitted for the degree of Doctor of Philosophy in the Imperial College London. The research described herein was performed under the supervision of Professor Jianguo Lin in the Department of Mechanical Engineering, Imperial College London, between April 2013 and March 2017.

I hereby certify that, to the best of my knowledge and belief, all the work presented here is original, except any other published or unpublished works from which acknowledgements and references have been given fully in the text and in the reference list.

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Acknowledgements

The author expresses the deepest gratitude to Professor Jianguo Lin for his ongoing supervision, guidance, encouragement, and support throughout the course of this study, and to Dr. Catrin M. Davies for the enthusiasm and provision of knowledge in this field. I would also like to extend my gratitude to Dr. Zhusheng Shi for his contributions of time and ideas to make my Ph.D. experience more meaningful and productive.

I am also grateful to my colleagues in the Mechanics of Material group, particularly Junyi Lee, Yong Li, and Qi Rong for their support in numerical computation.

This work is supported by the ESI Group in France and Embraer Company in Brazil. I am indebted to them for co-funding this project.

Special thanks are due for the dedication and skills of the technical staff of the Department, for the maintenance of equipment and for technical advice, especially Mr. Alex Toth for the laboratory management and Mr. Suresh Viswanathan Chettiar for the mechanical testing and forming.

Much credit is due to my parents and my siblings for their consideration, unconditional love, and encouragement.

Abstract

Creep age forming is a novel deformation process which applies creep/stress relaxation deformation on the age hardenable alloys for manufacturing large aluminium panels with complex curvature for modern vehicle and aerospace structures. One of the main issues that influence the accuracy of the forming shape in creep age forming is springback. Despite tremendous effort to predict the shape and springback of formed parts, springback has been analysed in terms of the elastic unloading of residual stress from stress relaxation. Anelastic creep strain recovery, however, may also substantially influence the springback and final shape of formed parts.

The present work was focused on evaluating the extent to which anelasticity can affect springback in the creep age forming process for the aluminium alloy AA7050-TAF. The mechanical properties of this material were characterized first with respect to age hardening and anelastic/inelastic creep deformation behaviour. Transmission electron microscopy (TEM) was used to examine microstructural variations of the material under the pure aged and creep aged conditions.

The Uniaxial Creep-Aging and Recovery Test (CART) and the Creep-Aging and Back Stress Test (CABT) were developed and performed to investigate the anelasticity and internal stresses induced during the forming process. The three-point bending Stress Relaxation and Back Stress Test (SRBT) was performed to examine the relationship between the thickness of the material and the amount of anelastic creep strain recovery after creep age forming.

A constitutive model was then developed to represent and predict the response of the material in creep age forming. The highlight of this research is that anelastic creep strain recovery has been proposed to influence the springback after the forming process; the constitutive equations allow the estimation of anelastic strain and other state variables from the creep aging histories of the material.

Keywords: Creep age forming; Anelasticity; Springback

PREFACE

What makes metal forming intriguing is that in even a simple forming process, the variations in the mechanical behaviour and microstructure of a material can be very complicated. An ultimate dream for a scientist conducting research on metal forming is to explore the most effective way to describe and predict the mechanical behaviours of materials during the forming process. To this end, numerous disciplines – mechanics, mathematics, physics, chemistry, and materials science – have been applied over the past few decades. Beyond these academic modalities, many other approaches, including trial and error experimentation, industrial engineering methods, and Taguchi methods, have been applied to expand and refine the predictions in metal forming.

At present, advanced efforts are still needed to fully expound the deformation mechanisms of materials during the forming process, and to clearly describe the mechanical behaviours and microstructural variations through material modelling. This problem is complex because it requires knowledge of both multiple mechanisms and their interactions for its solution.

Throughout considerable efforts expended on investigating creep age forming, the goal of my research was originally to investigate the creep aging behaviour of the aluminium alloy AA7050-TAF and to develop a set of unified creep aging constitutive equations for an over-aging condition of the material. As the research went on, however, many questions arose in my mind. The main question was, “Why shouldn’t we focus on the total deformation after

unloading from creep aging?” The core reason why this question occurred to me is that creep age forming is related to “manufacturing analysis”, in which the main goal is to predict deformation response – the final shape of the product after forming. In contrast, conventional studies in creep are more concerned with “structural analysis”, which focuses on creep and its damage mechanisms for improving creep strength, reliability, durability, prediction of failure, and operational strategies.

Generally speaking, the material’s deformation behaviours pertinent to creep are governed by three principal physical processes: elastic deformation, non-elastic deformation, and anelastic deformation. Previous studies in the field of creep age forming have provided great progress in the second category, providing formulated equations for predicting a material’s response to creep aging under most conditions.

Those previous works, however, generally treated creep as an “irreversible” non-elastic deformation. In fact, creep deformation exhibits components of both “irreversible” permanent deformation and “reversible” anelastic deformation [1-4], as well as creep recovery [5-7] and anelasticity [8-10]. Indeed, only a very limited number of literature surveys have discussed the mechanisms of anelasticity and their influence on metal forming.

After discovering this aspect, I clearly saw a great opportunity to continue my research and explore the mechanisms in detail. Gaining greater insight to unveil the awesome secrets of anelasticity is certainly a challenge. In this thesis, experimental studies – including newly designed testing methods – and modelling works are proposed. This research will hopefully contribute towards greater understanding of the mechanisms of anelasticity and offer pertinent guidance to industry that can greatly improve the precision of creep age forming.

Yo-Lun Yang

London, June 2017

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List of Nomenclature and Abbreviations

T	temperature
t	time
d	average diameter of grains
d_p	thickness of the plate in three-point bending
D	maximum deflection
D_{unload}	maximum deflection after unloading
D_r	maximum deflection after unloading
E	Young's modulus
k	averaging factor
ε_c	creep strain
$\dot{\varepsilon}_c, \dot{\varepsilon}_{ij}^c$	creep strain rate
$\dot{\varepsilon}_{\min}$	minimum creep rate,
$\varepsilon_v, \varepsilon_p, \varepsilon_{an}$	viscous strain, permanent strain, anelastic strain
$\varepsilon_e, \varepsilon_t$	elastic strain, and the total strain
σ_u	unit stress (1 MPa)
σ, σ_{ij}	applied stress in creep
σ_y	yield strength of the material
σ_A	strength contribution from age hardening
σ_{UTS}	ultimate tensile strength of the material
σ_{ss}	contribution of strength from solid solution hardening
σ_{dis}	contribution of strength from dislocation hardening
σ_b, σ_{ij}^b	back stress
σ_{int}	internal strength of the material
r, r_c	precipitate radius, and its value at peak-aged condition
ρ, ρ_i	dislocation density, and immobile dislocation density
ρ_m, ρ_{ij}^m	mobile dislocation density
ρ_{ini}	initial dislocation density
ρ_{max}	saturated dislocation density
$\dot{\rho}_m$	evolution rate of normalised mobile dislocation density
S_{ij}	deviatoric stress tensor
σ_{ij}^{eff}	effective stress tensor, $\sigma_{ij}^{eff} = \sigma_{ij} - n_0 \sigma_{ij}^b$
δ_{ij}	Kronecker Delta Function, $\delta_{ij} \begin{cases} = 1 & \text{if } i = j \\ = 0 & \text{otherwise} \end{cases}$
T_{ij}	mobile dislocation state variable
L	length of the material
F	applied load
F_r	load recovered during back stress stage
F_{end}	residual load before unloading
R_{load}	load relaxation ratio
$R_{\text{deflection}}$	percentage of deflection recovered ($(D_{\text{unload}} - D_r)/D_{\text{unload}}$)
SP_{instant}	instant springback, $(D - D_{\text{unload}})/D$
x	displacement in grain boundary sliding
$X, \bar{X}, \dot{X}$	variable X , normalised X , time derivative of X

TEM	Transmission Electron Microscopy
EDX	Scanning Electron Microscopy
CART	Creep aging and recovery test
CABT	Creep aging and back stress test
SRBT	three-point bending stress relaxation and back stress test
CPE4R	Four-node plane strain element in ABAQUS

1 INTRODUCTION

Creep age forming is a time-dependent inelastic forming technique that applies creep/relaxation at an aging treatment temperature. Creep is the time-dependent inelastic deformation that occurs under constant load condition, whereas relaxation is the time-dependent stress drop under constant deformation. In the formed parts, the forming process is accompanied by the evolution of microstructures, including the solid solution of alloying elements, dislocation movement, and precipitation hardening. The term “inelastic” is defined as time dependent permanent deformation and “plastic” is defined as time independent permanent deformation in this study.

The gradual change in microstructures not only affects the mechanical properties but also influences the creep strain recovery, or anelasticity, of the formed parts. The goal of this study is to develop a constitutive modelling technique for the prediction of creep recovery after forming.

This introductory chapter aims to present the general background to the work presented in this thesis, which includes aluminium alloys in use, the development of creep-age forming, the role of anelasticity in creep age forming, motivations and objectives for the research, and finally the arrangement of the thesis.

1.1 Introduction to aluminium alloys and their role in creep age forming

Technological advances in the automotive manufacturing, shipbuilding, and aerospace industries demand the use of strong, light materials. The most feasible materials for such applications are presumably aluminium alloys. Aluminium alloys possess many attractive properties [11, 12]. First, they are light, having an atomic weight of 26.98 and a density of 2.7 g/cm^3 , which is roughly one-third the weight of other commonly-used ferrous alloys. They also have a good strength to weight ratio, high corrosion resistance, easy workability, and high electrical and heat conductivity. This extraordinary combination of material characteristics makes aluminium alloys widely applicable in industry. Most notably, the applications of aluminium alloys have been increasing in the automotive [13] and aerospace [14] industries. As the applications of aluminium alloys grow in number, the industries using them are seeking new ways to improve the mechanical properties of the materials to reduce their weight. Apart from selecting new alloys with better material properties, an effective way to achieve high quality, reliable products is to find the optimum forming process for the metals.

The strength of a metal can be increased by solid solution hardening, age (or precipitation) hardening, and work hardening (dislocation hardening). These strengthening mechanisms are influenced by the metal forming history, including mechanical or thermo-mechanical treatment, heat treatment, and age heat treatment. Different heat treatment processes can greatly influence the corresponding microstructure and thereby affect its mechanical properties. Thermo-mechanical treatments are applied to form a desired shape, to produce a fine grain structure, and to promote dislocations so as to increase the strength of the metal. In general, the higher the amount of mechanical treatment prior to solution heat treatment, the finer the grain structure that can be achieved [15]. Solution heat treatment is a treatment to dissolve all the precipitates and second phase into the solid solution so that it becomes one single phase. Age heat treatment is a process to promote and control the growth of precipitates to improve the mechanical properties of materials.

For aluminium alloys, the most important strengthening mechanism is age hardening [16]. For example, the ultimate tensile strength of aluminium alloy AA7050 is around 200 MPa in the annealed state, but with proper aging heat treatment, the tensile strength can be

increased to 600MPa [17-19]. In order to age harden aluminium products, conventional metal forming techniques often require extra steps of aging heat treatment to improve the material properties. However, the additional processing can increase the costs and difficulty of the manufacturing process.

Creep age forming possesses special features that make it a feasible method of forming aluminium products. When creep age forming is performed at a certain elevated temperature to cause stress relaxation, age hardening can also occur concurrently in heat treatable alloys, especially aluminium alloys. This technology combines forming and aging treatment into a single process. With the proper forming conditions, the desired shape and mechanical properties can be achieved at the same time. The following section will provide a general introduction to creep age forming.

1.2 Advances in creep age forming technologies

Over the last few decades, significant progress has been made in developing new forming techniques in manufacturing technology [20-23]. Driven by the increasing requirements of various industries, new developments in this field are focused on the pursuit of materials with better performance, but environmental reasons are also important. Improvements to forming techniques allow products to be manufactured with less energy, and they also result in products with lower weight and better mechanical properties – better strength, better toughness, and higher resistance to fatigue and corrosion.

Creep age forming, under these premises, has been developed for manufacturing large aluminium panels with complex curvature for modern vehicle and aerospace structures such as aircraft wing skin panels and stringers [24, 25]. Such products are sometimes not easy to fabricate, and their fabrication usually requires more complicated steps in conventional forming methods such as roll forming, shot peen forming, brake forming, and stretch forming [25, 26]. These forming techniques, in general, are cold working or hot working processes that achieve the desirable shape by straining the metal beyond the elastic limit at an ambient or elevated temperature. For example, in stretch forming process, the metals are deformed plastically under combined bending and tensile stresses. The magnitude of the amount of plastic deformation is proportional to the force applied. Such forming process may result in non-uniformed strain distribution along the cross

section of the formed parts where some regions of the formed parts are not sufficiently deformed but other regions are deformed severely. This leads to heterogeneous dislocations and grain structure of material. The level of residual stress is, therefore, high.

In contrast, creep age forming technology is based on using creep/stress relaxation to deform sheet metal at a certain artificial aging temperature where the parts can be more uniformly deformed. Furthermore, the strength and other mechanical properties, such as the fatigue resistance and stress corrosion resistance of the material, can be improved by artificial age hardening. In creep age forming, stress relaxation and artificial aging can occur simultaneously. In fact, several steps of the process of conventional forming and heat treatment – can be achieved in a single step of the creep age forming process.

Since the creep age forming condition is strongly related to artificial aging, the initial temper condition of the material must be considered. Generally speaking, a material that has been solution heat treated is preferred, for in that state, no precipitates have developed. Solution heat treatment involves heating and holding the material at a suitable temperature for enough time to dissolve the precipitates and second phases. The alloying elements are uniformly distributed in the solid solution and the material becomes a single phase. Once in this state, the material should be cooled sufficiently rapidly that the alloying elements have no time to diffuse and thus remain in the matrix.

1.2.1 The forming process of creep age forming

The creep age forming process begins with elastic deformation of the part into a predetermined shape. In that condition, the forming part is then constrained by a clamping tool and moved to a furnace for further creep aging treatment. The time and temperature for this stage are determined by the artificial aging requirement. The elastic strain on the forming part is gradually transformed into inelastic strain with time and temperature due to stress relaxation creep/stress relaxation deformation. This is quite different from conventional forming techniques where plastic deformation is induced directly by applying force. Following the artificial aging treatment, the formed part is removed from the furnace and released from the constraining tool. In the stage of releasing the formed parts, whilst stress relaxation creates some permanent deformation, there is a tendency of the formed part to recover to the original shape. This tendency is known as springback. The cause of springback is that

the time and temperature are not long or high enough for the stress relaxation to fully transform the elastic strain into inelastic strain. There is, therefore, always some elastic strain that will ‘spring back’ at the unloading stage. A schematic graph of the creep age forming process is shown in Figure 1.1.

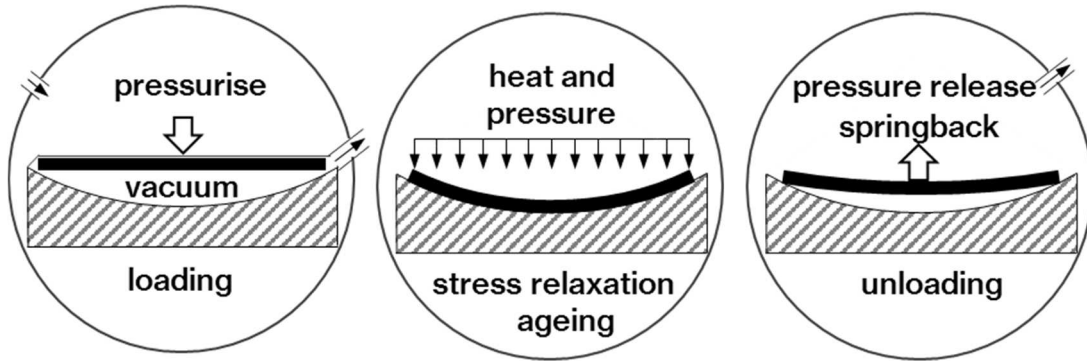


Figure 1.1. Creep age forming process in an autoclave.

In industrial practice, creep age forming is generally performed inside an autoclave, as shown in Figure 1.2. An autoclave is a computer-controlled pressurized furnace that can apply vacuum pressure on the lower surface of the forming part and high pressure on its upper surface. The vacuum pressure is generated by sealing the forming part with a vacuum bag and fitting the vacuum bag against the forming tool. As the air is removed from the bag, vacuum pressure between the lower surface of the forming part and the forming tool begins to rise. Meanwhile, the upper surface is subjected to high pressure generated by the autoclave. The pressure difference between the lower and upper surfaces of the forming part creates a bending stress on the forming part, which deforms the forming part such that it contacts the forming tool. Temperature is usually applied before pressure to soften the forming part and thereby decrease the forming pressure required to initiate the contact between the forming part and the forming tool.



Figure 1.2. Autoclaves for creep age forming [27].

1.2.2 Characteristics and future development of creep age forming

Products manufactured with creep age forming have been shown to have less residual stress than those manufactured with the conventional forming techniques [25, 28]. Residual stress is one of the key problems in metal forming [10, 25]. If not treated carefully, it may cause premature failure of the material [29]. The causes of residual stress are related to plastic deformation [30], the temperature gradient (uneven cooling rates of the outer and interior parts of the material) [31], and non-uniform structural changes due to phase transformation during the forming process. Conventional processes usually apply high magnitudes of plastic deformation over different areas of the product, so the residual stress can be high [25]. In creep age forming, on the other hand, the forming part is loaded with uniformly distributed stress on the surface to be contoured. The part does not need to be treated by large-scale plastic deformation. The temperature gradient and non-uniform structural changes are minor issues in creep age forming because the artificial aging temperature in creep age forming is lower than the temperature in conventional forming methods, and the parts are not subjected to severe structural changes.

One of the main issues that influence the accuracy of the forming shape in creep age forming is springback. The cause of springback is that the process time of creep age forming is governed not by the time required to form the material into the desired shape but by the time required for artificial aging to achieve the optimum mechanical properties of the material. The artificial aging time might not be sufficiently long for the material to be properly formed into the desired shape. In such a case, the springback of the product is greater than that of forming methods that apply plastic deformation.

The most common way for industries to overcome springback is by designing an appropriate forming tool [32]. First, a subscale forming trial is performed to evaluate the amount of springback on the forming part. An appropriate forming tool is then designed based on the springback compensation for the forming process. Sometimes more than one trial is required to design an applicable forming tool for use in mass production. This trial and error approach is not the best method because the costs of the forming tool and forming parts are high if the tools and parts are large. To reduce these costs, considerable research has been focused on computational methods to predict springback.

With advances in computer technology, computer-based numerical methods can be powerful tools in metal forming technology [33-35]. Such methods enable the prediction of not only the mechanical properties and final shape of the forming parts but also the springback [36]. Generally speaking, numerical methods require constitutive models to provide a material's mechanical properties for computation. The following section will introduce constitutive modelling. A detailed review of numerical methodologies for predicting springback will be provided in Chapter 2.

1.3 Constitutive modelling

With technological advances, computational methodologies have been developed for the purpose of reducing trials and errors of the forming processes. Of all these explorations, constitutive modelling has been found to be of great practical benefit [37]. Its efficacy in the metal forming process can be described as follows:

- Constitutive modelling can reduce the number of trial and error experiments needed for the prediction of new formed shapes.
- Constitutive modelling provides a more pragmatic description of the forming process, which avoids the complexity of theoretical explanations.
- The parameters of the forming process can be customized for different applications.

Constitutive equations are the medium by which a material's properties and behaviour acquired from theories and experiments can be put into a mathematical framework. They are an essential constituent (apart from other constituents such as equilibrium and compatibility equations) of reproducing or simulating the material's behaviour. The constitutive relationship of materials is determined by empirical descriptions based on experimental data for the same or similar conditions of the specific application.

The use of experiments for determining constitutive relationships are often simpler than using the specific applications. The testing range of various variables in the experiments is also broader than the applications. The reason is that the applications entail higher costs than simple experiments. Through the process of constitutive modelling, complex applications can be extrapolated from simple test data; e.g., from a limited number of test data to all of the stress levels and strain rates of interest, or from uniaxial tensile creep data to uniaxial compressive creep and a multiaxial creep age forming process.

Although the equations are determined from various experimental measurements under desirable experimental conditions, they cannot include all of the deformation sequences and the underlying physics that could affect the forming process. For example, forming a metal into a complex shape often involves non-proportional multiaxial deformation. Under this condition, the non-linear inelastic deformation response of materials becomes complicated. Even though much research has been conducted on microstructural characterization [38-40], the effects of the loading conditions, non-elastic deformations, and aging time on the internal structures (such as dislocation densities, precipitate evolutions, and internal stresses) remain unknown. Internal state variables can be introduced to overcome this difficulty [41, 42]. Microstructures such as dislocations, solid solutions, precipitates, and internal stresses all control inelastic deformation [43]. Using internal state variables to describe the microscopic properties can greatly simplify constitutive equations, facilitating their implementation in finite element (FE) software.

1.4 The role of anelasticity on springback in creep age forming

When a material is unloaded from creep, both the elastic strain and some of the creep strain induced during creep can be recovered with time. This phenomenon is described with a variety of terms: delayed elasticity, creep recovery, or anelasticity. It is defined as “a time-dependent reversible deformation” [3, 44], whereas elasticity is “a time-independent reversible deformation”. The role of anelasticity implies that when a formed part is released from the autoclave after the forming process, then apart from the springback related to elastic unloading, secondary time-dependent springback, or anelastic creep strain recovery will occur.

The phenomenon of time-dependent creep strain recovery accompanying the stress reduction or unloading from creep or stress relaxation has been well known for decades [45]. In the field of creep age forming technologies, however, this is still quite a new concept. Conventionally, springback, shape recovery, or other dimensional stability issues of the formed parts resulting from the removal of stress after the forming process have been analysed mainly in terms of the release of residual stress from creep or stress relaxation of the process. The material’s behaviour at this unloading stage is often treated as purely elastic, and the recovery rate is considered independent of time [24, 36, 46-48].

Under these premises, alloys in different tempering conditions have been extensively tested to build constitutive equations and thereby to predict the amount of springback [40, 49, 50]. Such approaches have been based on the consideration that the entire deformation of the material in the process is entirely irreversible. The anelastic creep strain recovery effect has long been believed to be insignificant – the anelastic strain is negligible as compared to the high permanent strain induced.

When the deformation is simply considered to be dominated by inelastic deformation or diffusional creep [51], such an explanation is well founded, since vacancy diffusion through the lattice leads to the generation/annihilation of vacancies, causing permanent strain but not recoverable anelastic strain. In this condition, the anelastic creep strain recovery effect does not take place.

Other creep mechanisms can lead to time-dependent anelastic recovery [52]; however, they may also play substantial roles in the creep age forming process [10, 53]. These mechanisms include the shear or sliding of grain boundaries, or dislocation rearrangement due to changes in internal stresses.

To the author's knowledge, previous studies in the field of creep age forming have not paid much attention to how the anelasticity could influence springback. As one of the main interests in this field is to predict the final shape of the product after forming, the influence of anelastic creep strain recovery on the formed part should be considered. One of the major objectives of this study is to characterise the anelastic effect through experimental methods. In the following chapters, related mechanisms will be proposed as the cause of anelasticity. How the anelastic effect could influence industrial applications will also be discussed.

1.5 The scope and aims of this study

Understanding the deformation and strengthening mechanisms of materials in creep age forming is crucial to the development of better constitutive models for FE simulation of industrial processes so that the optimum mechanical properties can be achieved.

The aim of this research was to investigate the age-hardening and creep behaviour of the aluminium alloy AA7050-TAF in order to understand the strengthening mechanisms of

this material that underlie the creep age forming process. Based on a series of investigations, a set of unified creep-aging constitutive equations is developed. These equations are further implemented into commercial Finite element ABAQUS via the user defined subroutine, VUMAT.

The goals of this study are listed below:

- To contribute towards greater understanding of the strengthening mechanisms behind age hardening and the influence of anelastic creep strain recovery, i.e., anelasticity, on creep deformation.
- To develop a more fundamental understanding of deformation mechanisms in creep age forming by applying different experimental techniques.
- To improve the precision of shape and springback prediction by introducing the concept of anelasticity into the creep age forming process.
- To propose a microstructural model of anelasticity and to develop a set of unified creep-aging constitutive equations based on the model.

1.6 Structure of Thesis

The thesis is divided into seven chapters. The present chapter provides a general introduction of aluminium alloys in use, forming technologies, constitutive modelling, and the role of anelasticity in creep age forming. By means of connecting all these topics together, the purpose of this study has also been highlighted.

Chapter 2 delves into the scientific backgrounds and recent literature of precipitation hardening, creep mechanisms in aluminium alloys, theories on anelasticity, and modelling techniques related to this study. Through this process, the underlying micro-mechanisms in creep age forming process can be characterised.

Chapter 3 presents all the experimental methodologies and programmes performed in this research. It includes an introduction of the equipment used, the fundamental mechanical testing methods, microstructural characterisations, and newly developed testing methods: the creep aging and recovery test (CART), the creep aging and back stress test (CABT), and the three-point bending stress relaxation and back stress test (SRBT).

Chapter 4 presents the experimental results acquired from the experiments of Chapter 3. An analysis and discussion of these experimental data is also provided in this chapter.

Chapter 5 introduces constitutive modelling techniques for applying the experimental results into a mathematical context. The first part of this chapter presents the modification of a set of unified constitutive equations developed by Zhan, the material constants of which are calibrated for the current material, aluminium alloy AA7050-TAF. In the second half of this chapter, on the basis of the literature review in Chapter 2 and the experimental studies in Chapter 4, a set of unified creep-ageing constitutive equations is developed for modelling the anelastic creep strain recovery effect. A theory of the principal micro-mechanisms in anelasticity is also proposed.

Chapter 6 first presents the method of implementing the newly-developed constitutive equations into the commercial FE software, ABAQUS. Then an integrated numerical procedure to simulate the creep age forming process is demonstrated. A FE simulation – the three-point bending stress relaxation and back stress test – is conducted. The simulation results are compared with the experimental results presented in Chapter 4.

Chapter 7 concludes by presenting the overall outcome and new findings of this research. Major contributions, including the newly designed testing methods and constitutive modelling techniques, are highlighted. This chapter also describes how the new findings and new testing methods can be applied into the manufacture process. Conclusions and suggestions are provided for industries requiring a better creep age forming process. Prospects and future work toward greater understanding in this field are also discussed.

2 LITERATURE REVIEW

This chapter reviews the theories and mechanisms behind the creep age forming process, including age hardening of materials, deformation mechanisms, and constitutive modelling. The role of anelastic creep recovery in creep deformation will be highlighted because it can strongly influence the shape of the formed parts. Creep deformation mechanisms will be only briefly presented, since the condition of a material in creep age forming is different from that in conventional creep and is beyond the scope of this research.

2.1 Principle of age hardening in Aluminium alloys

2.1.1 Aging heat treatment of Aluminium alloys

Aging heat treatment is often applied to aluminium alloys to improve the mechanical properties by precipitation hardening. Aging heat treatment begins after solution heat treatment – heating the alloy to a temperature below the lowest melting point and holding it for enough time to dissolve all the second phases. The purpose of solution heat treating aluminium alloys is to construct a solid-state supersaturated condition. The material is quenched to room temperature to retain the alloying elements in supersaturated solid solution.

The supersaturated condition state is unstable. The second phases with precipitates often naturally form, even at room temperature. The aging process can be accelerated by heat

treatment, an artificial aging process. For Al-Zn-Mg 7000 series alloys, precipitation often begins with the formation of Guinier–Preston zones, or GP zones, whose plane view of the habit plane is shown in Figure 2.1a [54]. Two types of GP zones can usually co-exist in the matrix, namely, GPI zones and GPII zones. GPI zones are single layers of metastable clusters containing high concentrations of Zn and Mg/Al atoms that are coherent with the matrix [55, 56]; the corresponding cross section views for edge-on habit plane configuration are shown in Figure 2.1b [57]. GPI zones can be formed in a temperature range from room temperature to 140-150°C, independent of the quenching temperature [58]. GPII zones are two layers of zinc-rich clusters sandwiching three layers of aluminium, as shown in Figure 2.1c [57]. They form after aging at above 70°C in Al-Zn-Mg 7000 solution treated at above 450°C [55, 58, 59].

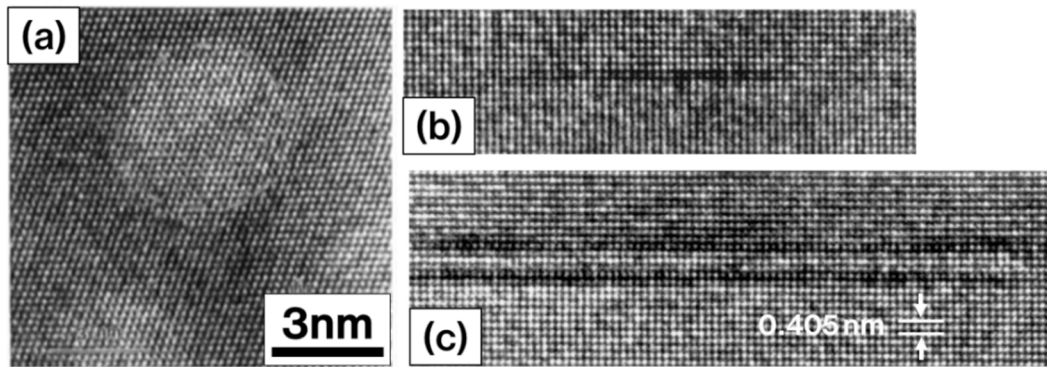


Figure 2.1. (a) A plane view (for habit plane) of a well-developed spherical coherent GP zone after artificial aging at 95°C for 20h[54]. Cross section views of (b) GPI and (c) GPII zones in Al-4wt%Cu taken along [100] zone axis of Al matrix [57].

During aging, both GPI and GPII zones transform into metastable hexagonal phase η' precipitates with a platelet-shaped morphology that is semi-coherent with the matrix [60-62]. Finally, the η' precipitate phase transforms into the equilibrium η precipitate phase (usually known as a Laves phase, MgZn_2) with plate-shaped or rod-like morphology [63], depending on the section of observation. The η precipitate phase is presumed to be incoherent with the matrix. Because the GP zones and the η' precipitate phase have a higher free energy than the η precipitate phase, the precipitation of the η phase easily form from the GP zones and η' phase, since they can overcome the small energy barriers due to the coherent interface boundary with the matrix [64]. The precipitation transformation sequence can be described as below [65]:

Solid solution (α) $\rightarrow$ GP zones $\rightarrow$ η' precipitates $\rightarrow$ η precipitates (MgZn_2)

Sometimes a single quenching and aging treatment cannot achieve the desired mechanical properties. For example, during a single aging treatment, precipitates of ideal size and density may form inside the grains; however, a precipitate-free zone (PFZ) forms along the grain boundaries and around the locations of efficient vacancy sinks. This heterogeneous microstructure decreases the mechanical properties significantly [66, 67]. To overcome the above difficulty, double aging treatments have been introduced [68]. In Al–Zn–Mg 7000 series alloys, it has been found that the double aging treatment leads to the formation of a finer precipitate structure, giving better properties. This treatment includes solution heat treatment and quenching, pre-aging, and second aging. The mechanism of double aging treatment is that the first aging treatment forms GP zones, and some GP zones start to transform into η' precipitates. The second aging accelerates the kinetics of transition of the GP zones to η' so as to reduce the time to achieve maximum strength, as shown in Figure 2.2. For double aging treatment, in order to accelerate the transformation of the GP zones to η' with copious nanometer-sized precipitates, the aging temperature must be controlled carefully.

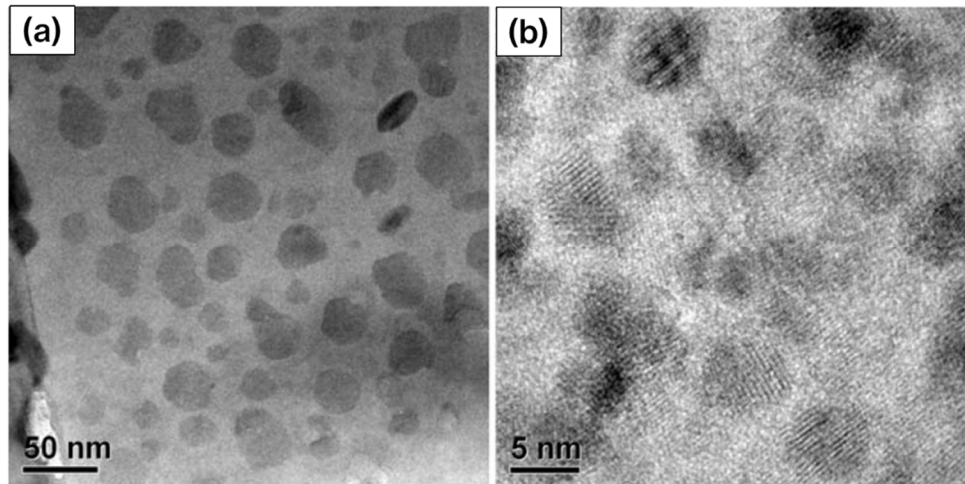


Figure 2.2. (a) TEM microstructure of single aging treatment at 177°C for 55 min with coarse hexagonal η' precipitates. (b) TEM microstructure of double aging treatment for 55 min at 121°C and 55 min at 177°C. TEM observation was along the plane view of habit plane of η' precipitates [69].

It has been reported that Al–Zn–Mg alloys heat-treated with the first aging temperature lower than the second aging temperature tend to be harder than those heat-treated at the same temperature in both steps [70, 71]. The main reason may be that the surface energy of GP zones is low [58], so it is more likely that at a low aging temperature, a high number density of GP zone structures or very fine transition phases will form, both of which are beneficial

for nucleation of η' precipitates [69]. If the aging temperature is high enough to form more stable phases such as η' precipitates, it is unlikely to dissolve them and form GP zones again at lower temperature in the second stage aging [72]. It is, therefore, appropriate to conclude that the first aging temperature should be lower than the second aging temperature in order to create highly dense GP zones in the matrix, thereby forming considerable amounts of nanometre-sized η' phase during the second aging. It has been discovered that in the creep age forming process, the stress relaxation rate of AA7475 alloy increases during the second aging step [73], presumably because the previously formed GP zones in the first step dissolve and nucleate metastable η' phase [74]. A high stress relaxation rate is beneficial for creep age forming because it can decrease the springback effect.

2.1.2 Strengthening mechanism of age hardening

The mechanism of age hardening in aluminium alloys is generally related to the interaction of precipitates and moving dislocations. The precipitates, or GP zones, with the strain field act as obstacles hindering the movement of dislocations [75, 76].

The sources of age hardening can be further analysed into at least six causes:

- (1) Coherency strain: The hardening is caused by slight misfit of the lattice parameter or orientation between the precipitates and the matrix. The misfit gives rise to elastic stresses around the precipitates, which impede the movement of dislocations.
- (2) Chemical hardening: When a precipitate is sheared by dislocations, the increasing interphase boundary around the precipitate increases the energy associated with the interface, resisting the process. Therefore, additional stress must be applied to the dislocation to force it through the precipitate.
- (3) Order hardening: When a dislocation cuts through an ordered precipitate, the destruction of atomic bonds increases the internal energy and creates anti-phase boundaries.
- (4) Stacking-fault: The hardening results from differences in the stacking fault energy between the precipitates and matrix, which increases flow stress because of the different separations of partial dislocations.

- (5) Modulus difference: The hardening occurs when the difference in the elastic modulus between the precipitates and matrix induces imaginary forces as a dislocation approaches a precipitate.
- (6) Dispersed particle: The hardening occurs when the alloy contains incoherent precipitates or particles. Because the dislocations need to bypass these obstacles, the shear stress of the alloys increases.

In Al-Zn-Mg 7000 series alloys, the key hardening constituents are the coherent GP zones and the semi-coherent η' precipitates [77]. In the initial stage of aging, the lattice distortion between GP zones and the matrix provides elastic stresses, resulting in coherency strain hardening, which impedes dislocations as they cut through. As a result, the strength of the alloys increases. When the GP zones transform into semi-coherent η' precipitates (5 – 20 nm), the lattice strain field around the η' precipitates and the η' precipitates themselves can impede the movement of dislocations, thereby strengthening the alloys.

Small-sized particles with coherent boundaries generally permit a dislocation to pass through and cut the particles. Conversely, dislocations tend to bow between large particles or between particles with disordered interphase boundaries. That is, the so-called dislocation bowing, proposed by Orowan (the so-called Orowan mechanism), occurs when the particles exceed a certain critical size or when the interphase boundary is disordered. The incremental contribution to the yield strength via precipitation hardening depends on whether dislocation “cutting” or “bowing” is the predominant slip mechanism.

In a later stage of aging in Al-Zn-Mg 7000 series alloys, the η' precipitates gradually transform into incoherent large-sized η precipitates (50 nm) with a low number density. Because the size and the incoherency of the η precipitates are great enough, the dislocations cannot cut through. The dislocation can move only by sharp changes in curvature of the dislocation line, bending between the precipitates and bowing around the particle. As a result, a dislocation loop forms around each precipitate as at a Frank-Read Source [78]. The dislocation can continue to move, but it leaves a dislocation loop surrounding the precipitate, the stress field of which increases the resistance to the motion of the next dislocation. In the over-aged condition, the η precipitates (particles) continue to grow and coarsen, and the particle spacing increases with the decrease in the number

density of precipitates. Consequently, the applied stress (under the Orowan mechanism) to move the dislocation decreases, as the yield stress is determined by the shear stress required to bow a dislocation line between two particles, which is inversely proportional to the particle spacing. The related formula can be expressed as $\tau = Gb/L$ [79], where τ is applied shear stress, G is shear modulus, and L is particle spacing. Figure 2.3 shows the mechanisms and the evolution of yield strength with aging time of quenched Al-4wt% Cu alloy [80].

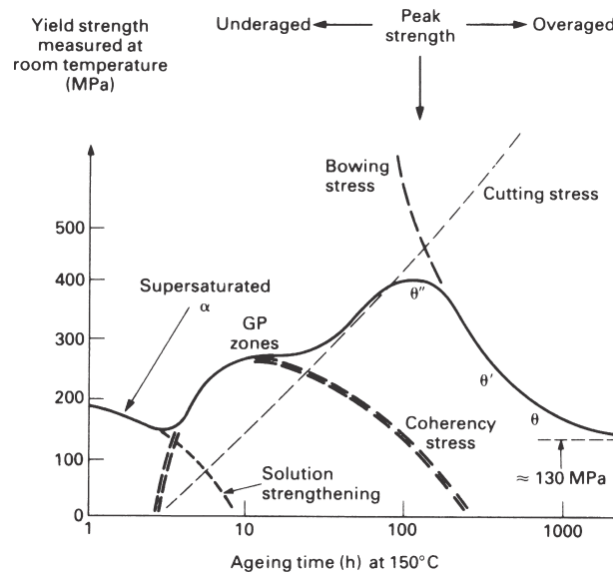


Figure 2.3. Yield strength evolution with aging time of quenched Al-4wt% Cu alloy [80].

The dislocation density of a creep aged alloy is higher than that of a pure aged one due to the effect of stress. The dislocations can become the nucleation sites of the precipitates. The related research works, however, have seldom dealt with the issue of how applying stress during aging process can influence the morphologies of the material and thereby affect its mechanical properties. By using mechanical testing and microstructural analysis methods, the present study will also investigate the influence of applied stress on the microstructure under the creep aging condition.

2.2 Creep deformation mechanism

Creep of materials is a time-dependent inelastic deformation under a constant load condition. The deformation process can be due to different types of defects of materials [81], namely, point defects (diffusional creep), line defects (dislocation creep), or planar defects (grain boundary sliding), or a combination of these three effects. The creep

deformation rate generally increases with temperature, since these defects in the material are more likely to occur at temperatures higher than $0.4-0.5T_m$, where T_m is the absolute melting temperature [82].

In the creep age forming process, unlike in conventional creep deformation, the operation temperature is relatively low, about $0.3-0.4T_m$; the time is much shorter, roughly 8 - 72h; and the applied load is high, sometimes even above the high temperature yield point of the material [21, 24, 26, 38, 83, 84]. The forming temperature and time are determined by the required artificial aging temperature and time for the material. The load is determined by the load required to form the shape of the formed part.

Another characteristic that distinguishes the creep age forming process from conventional creep is that the microstructural evolutions, especially the precipitate evolution, of the material are very fast, whereas conventional creep is generally investigated under conditions of constant internal structure [85-88]. Furthermore, the creep age forming process is not associated with deformation under a constant load of creep but under the stress relaxation deformation condition.

All these factors increase the complexity of identifying the deformation mechanism of the material. However, it can be generally evaluated with the deformation-mechanism maps developed by Ashby [89], which show that the deformation mechanisms underlying the creep age forming process can be time-dependent inelastic deformation due to dislocation glide, or power-law creep.

2.3 Anelasticity

Anelasticity, or creep recovery of materials, is a time dependent recovery process of strain when the applied stress is decreased or removed. Descriptions of anelasticity are often based on the idea that the material itself possesses an opposing stress that resists creep deformation when it is under an applied stress. When the applied stress is removed, the opposing stress works on the material, which causes anelasticity. The terms *threshold stress*, *long range internal stress*, and *back stress* have all been used for the concept of internal stress.

The term *back stress* is adopted in this research because it can pertinently describe the stress that causes the backflow of strain in anelastic strain recovery. Many micro-mechanisms

have been proposed to explain the possible mechanisms underlying anelasticity. They can be generalised into two metallurgical terms: planar defects regarding the grain boundary effect, and line defects related to dislocation motions. Point defects such as vacancies are not considered because this kind of defect is more closely related to the diffusional creep that mainly occurs at high temperatures ($>0.5T_m$) [90].

2.3.1 The effect of grain boundary shearing and movement on anelasticity

Creep deformation of polycrystalline materials often involves grain boundary shearing and sliding. For stress relaxation and anelasticity, the interactions between grain boundaries can be expressed as a two-component system, as shown in Figure 2.4 [45]. Figure 2.4a shows the material in the fully stress-free condition. The region within the rectangle is considered to be viscous, and the region outside of the rectangle, purely elastic. When the material is fully loaded, both regions are uniformly deformed, as shown in Figure 2.4b. As the material being stressed relaxed, the interior region is gradually deformed and the exterior region remains rigid. In order to maintain the continuity of the material, the surrounding exterior region has to undergo at least partial relaxation, as shown in Figure 2.4c. When the load is removed, the instantaneous recovered strain that accompanies the removal is uniform throughout. The part of the exterior region that is far away from the interior region is fully elastically recovered and stress free. Around the boundary of the interior region, however, the exterior region has residual stresses due to the anelastic strain remaining in the interior region, as shown in Figure 2.4d. A viscous process of anelastic creep strain recovery then occurs to rebalance the internal stress and anelastic strain so as to return to the initial state in Figure 2.4a.

A remarkable feature of the two-component system is that stress concentrations can be expected at the elastic and viscous boundaries. Raj and Ashby [91] proposed that the accommodation of internal stress in grain boundaries could be due to diffusional creep and dislocation motions. Elastic accommodation forces were proposed to be the cause of grain boundary sliding back stress [92]. Wei et al. [93] developed a model for the recoverable creep deformation of nano-crystalline Al and gold film [94] based on the stress concentration between the grain boundaries due to heterogeneous grain boundary

diffusion and sliding. Back stress is proposed to be dependent on the deformation history and current stress or strain rate conditions [9, 95].

Inelastic deformation of polycrystalline materials requires compatibility during the deformation of neighbouring grains [96, 97]. It is suggested that when a metal is under an applied stress, the strain incompatibility between grains and the substructure can cause internal stress [98]. These non-uniform distributed internal stresses can affect dislocation motion and cause anelasticity.

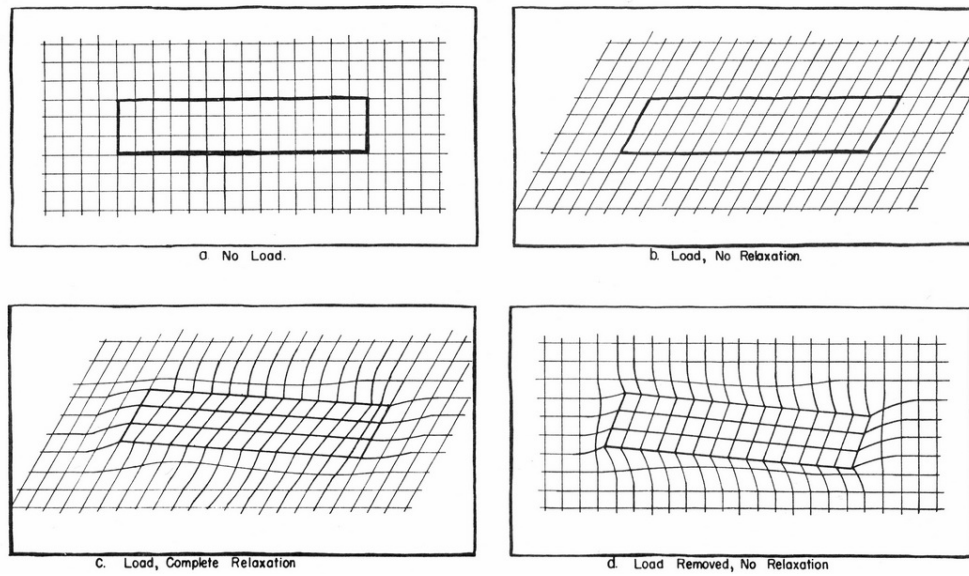


Figure 2.4. Illustration of the stress concentrations of material interfaces introduced by stress relaxation before loading (a), after elastic loading (b), under the state of stress relaxation with stress concentration between the interfaces (c), and after removing the load from stress relaxation (d) [45].

Sinha proposed a micro-mechanism [99] and developed the strain relaxation and recovery test for investigating anelastic strain recovery in the primary creep stage of a polycrystalline material [3, 4, 52, 100-103]. Anelasticity has been proposed to be associated with the reversible part of the interactions between interfaces, such as grain boundaries and precipitate-matrix boundaries [1]. In the primary creep stage, the deformation of a polycrystalline material involves shearing or sliding of the grain boundary region [99]. Grain boundary movement induces internal stress concentrations or elastic distortions of the lattice near the grain boundaries, ledges, and triple points or boundaries, as shown in Figure 2.5. “ d ” is the average diameter of grains, x is the displacement in grain boundary sliding, and k is the averaging factor, $\epsilon_{an} = \epsilon_{gbs} = Kx/d$ [1], where ϵ_{an} is the anelastic strain and ϵ_{gbs} is the strain caused by grain boundary sliding.

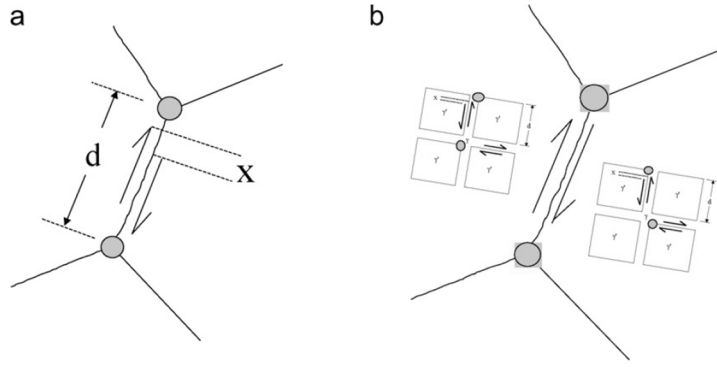


Figure 2.5. (a) Schematics for intergranular boundary shearing causing stress concentrations at triple points and (b) intragranular interface shearing resulting in “minor” stress concentrations in a triangular precipitate [99].

Since the recoverable anelastic strain is observed to increase during primary creep [9], primary creep can be seen as the summation of anelastic strain and inelastic strain. The viscous strain rate was discovered to be constant, and the anelastic strain followed a logarithmic decay function of time in primary creep [2].

Based on this concept, total creep strain can be divided into three components: elastic strain, ε_e , time dependent reversible (anelastic) strain, ε_{an} , and irreversible (viscous) strain, ε_v . Sinha [1] proposed a non-linear visco-elastic model to describe primary and steady state creep behaviour where the anelasticity is due to the elastically stored energy in the material: time dependent reversible distortions of intra-granular and inter-phase regions by shearing of these boundaries [52]. The total strain in creep can be expressed as follows:

$$\varepsilon_T = \varepsilon_e + \varepsilon_{an} + \varepsilon_v$$

2.3.2 The role of dislocation motion on anelasticity

A unified description of anelasticity has been proposed to incorporate dislocation motions [104]. Gibeling et al. [9] studied the anelasticity of polycrystalline and single-crystal pure Al, Cu, and Pb metals. Since the role of grain boundaries can be neglected in single crystals, it is therefore suggested that the physical mechanisms that contribute to creep anelasticity involve the motions of dislocations. Climb of dislocations was proposed to be the cause of anelasticity [5, 105].

Anelasticity has been discussed in connection with long-range internal stress, or back stress [106-109]. Back stress is the deviation of the effective stress from the applied stress that occurs over the micrometre range, such as in grains and dislocation cells. It has been suggested that it is generated in crystalline materials after plastic deformation [92, 110-113], cyclic deformation [97, 114-116], and creep [117, 118].

Geometrically necessary dislocations (GNDs), generated in inelastic deformation due to misorientation of lattice planes or grains, was proposed to give rise to localised back stress [113]. Back stress has also been attributed to heterogeneous dislocation sub-structures [119-122], involving the process of straightening of the bowed dislocation links and the formation of dislocation cell structures under the applied stress, and the reverse motion of dislocations in the dislocation cell walls.

An illustration of simplified two-dimensional dislocation cell structures is shown in Figure 2.6. The grids in Figure 2.6a represent the two sets of slip planes of the material. Dislocations pile up at the exterior of cell walls as the cell walls obstruct dislocation glides. The dislocation cell walls provide a back stress that has the same sign as the applied stress, whereas the cell interiors provide a back stress that has an opposite sign of the applied stress, as shown in Figure 2.6b. Such dislocation cell structures are believed to form during primary creep due to heterogenization of the spatial dislocations [119, 123]. Steady state creep is generally believed to start as soon as the dislocation cell structures are fully developed. Back stress has been demonstrated to be independent of the creep strain generated in creep deformation [124].

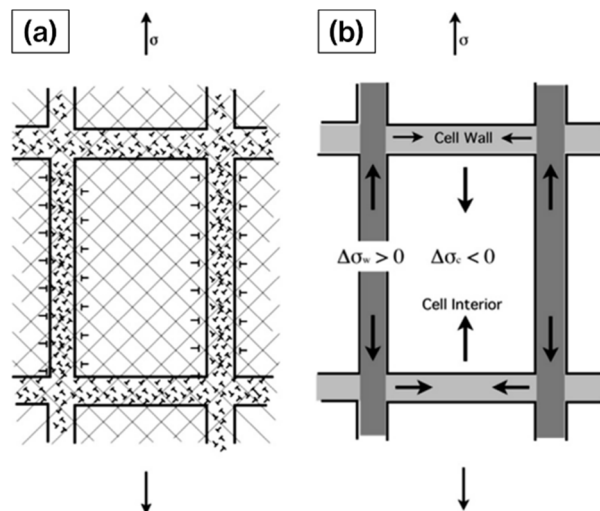


Figure 2.6. An illustration of dislocation cell structures showing the heterogeneous back stress distribution along the cell walls and cell interior [106].

Figure 2.7 presents a model of soft and hard regions of dislocation cell structures based on their associations with a localised internal stress field [120, 121, 125, 126]. When the material is under an applied stress, σ_a , dislocations gradually pile up and form a cell wall. Shear stress is then generated on the cell wall, since the dislocations cannot easily pass by it. On the other hand, compatibility requires dislocations to flow within the cell in an opposite sign. This leads to back stress of the cell interior, σ_b^{ci} , that has the opposite sign of the applied stress, and a back stress that has the same sign as the applied stress on the cell wall, σ_b^{cw} . At this stage, the total effective stress on the cell wall is $\sigma_b^{cw} + \sigma_a$, and the stress of the cell interior is $\sigma_b^{ci} + \sigma_a$. When the applied stress, σ_a , is removed, the existence of back stress in the cell interior, σ_b^{ci} , causes the backflow of the strain, [95]. The average value of back stress on the cell wall and cell interior can be expressed as $\sigma_b = (\sigma_a - \sigma_b^{ci}) = (\sigma_b^{cw} - \sigma_b^{ci}) \frac{f_{cw}}{f_{ci}}$, where f_{cw} and f_{ci} are the areal fractions of cell wall and cell interior ($f_{cw} + f_{ci} = 1$).

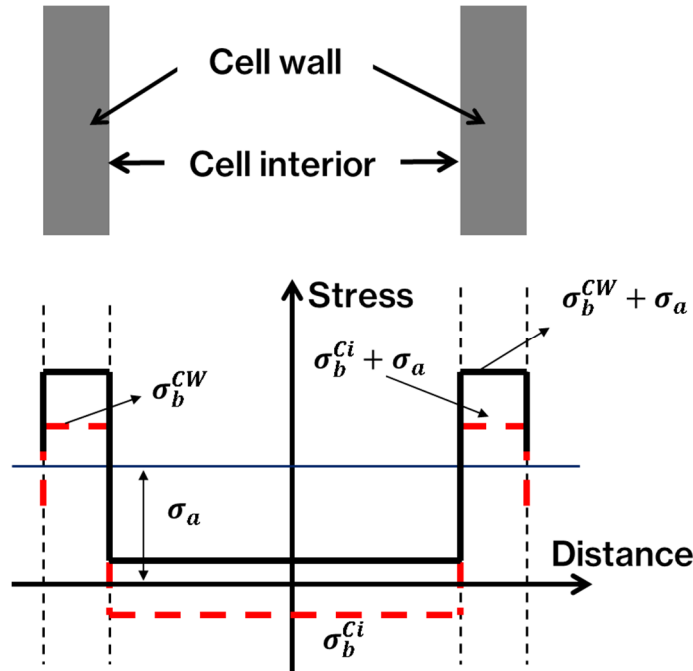


Figure 2.7. A schematic representation of the back stress model of a dislocation cell structure showing the local stress distribution across cells and cell walls, where the solid line illustrates the back stress, σ_b , under an applied stress, σ_a , and the dash line represent the back stress after removing the applied stress [119, 127].

2.3.3 Measurement of back stress: Mechanical testing methods

Since the back stress is generated by the dislocation cell structures, which are dependent on the applied stress condition, they can be measured when the applied stress changes.

Based on this idea, various mechanical testing methods have been developed in an attempt to measure the back stress, including the stress relaxation test [128], incremental unloading test technique [129], strain transient dip test [5, 130], and stress transient dip test [131-133]. All these tests are based on these assumptions [127]:

- The material is in a constant microstructure condition over the testing period, which involves rapid changes in applied stress, so the measured back stress can reflect the microstructural condition of the material.
- Anelastic creep recovery due to the back flow of dislocations, driven by back stress, can compensate for the dislocations that glide forward, driven by the effective applied stress; when the applied stress is reduced to a certain critical value where strain rate $\dot{\epsilon} = 0$, or stress rate $\dot{\sigma} = 0$, the applied stress can represent the back stress [9].
- The localised internal stress of dislocation cell structures is periodic and has an average value. The back stress is used to represent the average value of internal stress over the microstructure.

The stress relaxation test provides the simplest way to measure internal stress. However, this test takes such a long time that the assumption of constant structure might not be valid [134]. The incremental unloading test was developed to replace the stress relaxation test so that the time for stress relaxation could be shortened. In this test, the applied stress is incrementally decreased after a designated time and the strain is held as constant under each certain stress level for the stress relaxation process. These steps end when no more stress relaxation and only stress recovery occurs at a certain level of applied stress, as shown in Figure 2.8a, b. The internal stress determined from these tests, however, is difficult to interpret as the result of the back stress, which is driven by dislocation back flow and internal resistance, or mechanical threshold stress, below which no deformation will occur, or the combination of these two effects.

For strain transient dip creep tests, the applied stress is reduced and held constant at that level. The corresponding strain variation is recorded. The back stress is determined by incrementally reducing the applied stress until the creep strain rate becomes zero. If the applied stress is reduced so much that the back stress is higher than the applied stress, then the effective stress becomes negative and a backflow in strain can be expected, as shown in Figure 2.8c, d. This testing method is based on the consideration that the dislocation

velocity is controlled by the net stress of applied stress and back stress [5, 95]. When the creep strain rate becomes zero, the velocity of dislocations is zero and the average back stress is equal to the applied stress.

Figure 2.8e, f shows an illustration of the stress transient dip test. The test begins with reducing the applied stress to a certain level, where the strain of the sample is held constant for stress relaxation, and the stress variations are monitored. If stress relaxation occurs, the reduced applied stress is higher than the back stress. But if stress recovery occurs, the back stress is higher than the reduced applied stress. The back stress is determined by reducing the applied stress to a level at which no more stress relaxation or stress recovery will occur.

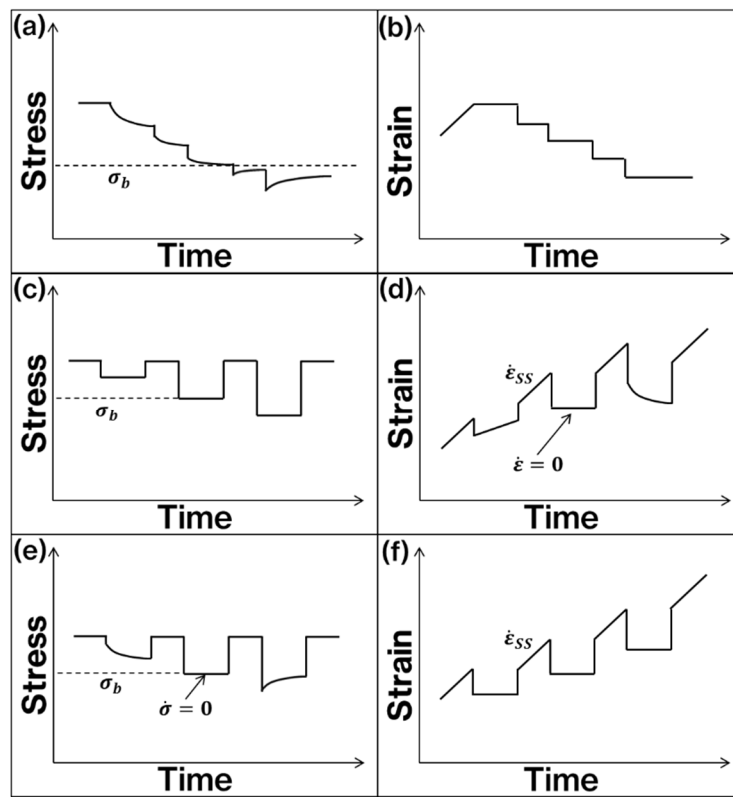


Figure 2.8. Schematic illustration of stress change testing techniques for determining back stress. (a), (b) Incremental unloading test, (c), (d) strain transient dip test, and (e), (f) stress transient dip test.

2.4 Unified Constitutive models for Creep age forming

For the past few decades, efforts have been made to develop mathematical models to predict the mechanical properties, form shape, and springback in the creep age forming process so that parts with a specified strength and shape can be achieved [36, 135]. These models are generally expressed in constitutive equations, which describe the state of

materials during the forming process. This review aims to present the key development of state variable theory and the origin and evolution of the constitutive equations that are used in this research.

2.4.1 State variable approach

The development of state variable theory began with Hart [136-138], who first proposed the rate equation of state in terms of stress, inelastic strain rate, and hardness evolution rate, in which hardness evolution rate is treated as a state variable that characterises the structural state of the material during the deformation process. The evolutionary equations proposed in this theory provide a solution for dealing with time dependent strain response. The central equation for the inelastic strain is in rate form, in which time independent plastic strain and time dependent creep strain are combined into the total inelastic strain rate ($\dot{\epsilon}$), and $\dot{\epsilon}$ is dependent on other states or structural variables which are also governed by rate equations, $\dot{X}$. The general form can be expressed as follows [139]:

$$\dot{\epsilon} = f(\sigma, T, X)$$

$$\dot{X} = g(\dot{\epsilon}, T, X)$$

Based on state variable theory, extensive experimental works were performed to develop sets of constitutive equations to describe the various phenomena occurring in plastic deformation. Anelasticity and plastic deformation were considered in Hart's original model. Alexopoulos et al. [140] improved the model for microplastic strain and recovery. Other constitutive equations also treat the Bauschinger effect [141] and grain boundary sliding [140, 141] as a form of state variable. With the use of state variables, these constitutive equations have successfully described and predicted various phenomena during deformation [142].

Krieg et al. [143] developed a constitutive creep plasticity model for rate-dependent deformation of metals. The model introduced two internal variables to represent hardening and recovery variables for the Bailey-Orowan process. The internal variables were incorporated with inelastic strain rate and applied stress in a power law relationship. These internal variables cannot be observed from experiments but are presumed to

provide corresponding information for the behaviours of material states that are left unquantified [144, 145].

MATMOD (MATERials MODel) equations were developed to model the behaviour of materials during creep or plastic deformation [139]. The evolution of the microstructure during deformation is expressed in the form of variables representing certain microstructural features or deformation mechanisms. This model, in general, is controlled by three types of structural variables: isotropic work hardening, directional work hardening, and intrinsic strengthening [146]. The model covers a very wide range of phenomena, including strain-rate and temperature dependent plasticity and creep; cyclic deformation of the Bauschinger effect and cyclic hardening and softening [147, 148]; static and dynamic recovery and the effect of stress on recovery; complex histories involving temperature, stress, and strain-rate changes during deformation; and multiaxial deformation [149, 150].

For the age hardening effect, a process model was developed [151]. The volume fraction of precipitates and the average particle radius are described by internal state variables, and the coarsening and refinement of the precipitates are described by the average radius growth as a function of time. A mathematical model for creep age forming [152] was first proposed in 1991 for cylindrical mould shapes, and the predicted data were consistent with experimental data from Textron, Inc. [25].

Many approaches have been used to construct constitutive models based on different materials, mechanisms, and experiments [20, 35, 37, 153-156]. These models therefore have their own strengths in different fields. Microstructure studies, especially with the use of transmission electron microscopy (TEM), can provide great insight into the micro-mechanisms of the materials during forming. For constitutive modelling, however, efforts still need to be made to quantify the TEM results to contribute directly to formulating the equations, since most constitutive equations include some empiricism and approximations.

2.4.2 Review of constitutive models for the present study

The constitutive equations used and developed in the present study originated from Kowalewski [157], who in 1994 proposed creep damage constitutive models for aluminium

alloys. This model uses three internal state variables – dislocation hardening, precipitate hardening, and damage variables – and two hyperbolic sine functions to describe the corresponding microstructural features and mechanisms during primary and secondary creep deformation. The equations are listed as follows:

$$\dot{\epsilon}_c = \left[\frac{A}{(1 - \omega_2)^n} \right] \sinh \left[\frac{B\sigma(1 - H)}{1 - \phi} \right] \quad (2.1)$$

$$\dot{H} = \left(\frac{h_c}{\sigma} \right) \dot{\epsilon}_c \left(1 - \frac{H}{H^*} \right) \quad (2.2)$$

$$\dot{\phi} = \left(\frac{K_c}{3} \right) (1 - \phi)^4 \quad (2.3)$$

$$\dot{\omega}_2 = D \dot{\epsilon}_c \quad (2.4)$$

where

$$n = \left[\frac{B\sigma(1 - H)}{1 - \phi} \right] \coth \left[\frac{B\sigma(1 - H)}{1 - \phi} \right] \quad (2.5)$$

The material constants A and B are related to secondary creep. H represents the dislocation hardening, which is related to primary creep. The initial value is 0 and increases as the creep aging process reaches H^* (the maximum value at the end of primary creep). Once it reaches the maximum, the value will remain the same for the rest of the creep process. ϕ is a state variable, which models the age hardening effect on the over-aged condition. ω_2 is a damage variable, which represents the damage at the tertiary creep. The time for the creep age forming process is far shorter than the creep failure time. The values of ϕ and ω_2 are therefore close to zero during the creep age forming process.

In 2007, Huang et al. removed the creep damage part of the equation from Kowalewski's equation for the creep age forming process as follows [158]:

$$\dot{\epsilon}_c = A \sinh[B(\sigma - \sigma_0)(1 - H)^{m_0}] \quad (2.6)$$

$$\dot{H} = \frac{h_c}{\sigma^{m_1}} \left(1 - \frac{H}{H^*} \right) \dot{\epsilon}_c \quad (2.7)$$

The above equations are based on high temperature creep theory and can only model simple creep deformation. The reason is that the creep age forming process is subject not only to creep but also to the age hardening effect, whereas this set of equations uses only

one state variable, $\dot{H}$, to model dislocation hardening without considering the interaction between creep and age hardening during the forming process.

In order to model the nucleation and growth of precipitates, and their effects on mechanical properties during the creep age forming process, Ho et al. [135] established a set of constitutive equations to model primary and secondary creep deformation with precipitate nucleation and growth for 7000-series Al-alloys. These constitutive equations are utilized in FEM modelling for thick plates with double curvature [47]. The equations are listed below:

$$\dot{\epsilon}_c = A \sinh \left[B [(\sigma - \sigma_A)(1 - H)] \left(\frac{C_{ss}}{\sigma_{ss}} \right)^n \right] \quad (2.8)$$

$$\dot{H} = \left(\frac{h_c}{\sigma^{0.1}} \right) \left(1 - \frac{H}{H^*} \right) \dot{\epsilon}_c \quad (2.9)$$

$$\dot{r}_p = C_0 (Q - r_p)^{m_0} \left(1 + \left| \frac{\dot{\epsilon}_c}{\gamma} \right|^{m_1} \right) \quad (2.10)$$

$$\sigma_A = C_A r_p^{m_2} \quad (2.11)$$

$$\sigma_{ss} = C_{ss} (1 + r_p)^{-m_3} \quad (2.12)$$

$$\sigma_y = \sigma_{ss} + \sigma_A \quad (2.13)$$

where r_p is the average spherical precipitate radius; its growth is controlled by the equation (2.10). σ_y represents the variation of yield stress, which changes during the creep age forming process. σ_A is the stress related to the contribution of precipitates during age hardening, and σ_{ss} is the stress related to solid solution hardening.

The material constants of the above constitutive equations were determined by creep aging experiments, where the creep deformation and precipitation hardening can be obtained by experimentation. The constitutive equations can model the mechanical properties, creep deformation, stress relaxation, and springback effect. However, these equations do not consider the change in volume fraction and the shape of precipitates. They have only limited ability to predict the creep strain, stress relaxation effect, precipitate hardening, and their interactions.

Zhan et al. proposed a set of mechanism-based unified creep aging constitutive equations including the evolution of precipitates and dislocation density, which control age hardening, solid solution hardening, and work hardening due to creep. The equations are listed below:

$$\dot{\epsilon}_c = A_1 \sinh\{B_1[|\sigma|(1 - \bar{\rho}) - k_0\sigma_y]\} \text{sign}\{\sigma\} \quad (2.14)$$

$$\sigma_y = \sigma_{ss} + \sqrt{\sigma_A^2 + \sigma_{dis}^2} \quad (2.15)$$

$$\dot{\sigma}_A = C_A \dot{\bar{r}}^{m_1} (1 - \bar{r}) \quad (2.16)$$

$$\dot{\sigma}_{ss} = C_{ss} \dot{\bar{r}}^{m_2} (\bar{r} - 1) \quad (2.17)$$

$$\dot{\sigma}_{dis} = A_2 \cdot n \cdot \bar{\rho}^{n-1} \dot{\bar{\rho}} \quad (2.18)$$

$$\dot{\bar{r}} = C_r (Q - \bar{r})^{m_3} \cdot (1 + \gamma_0 \bar{\rho}^{m_4}) \quad (2.19)$$

$$\dot{\bar{\rho}} = A_3 (1 - \bar{\rho}) |\dot{\epsilon}_c| - C_p \bar{\rho}^{m_5} \quad (2.20)$$

where the creep strain rate $\dot{\epsilon}_c$ is a function of stress σ , dislocation density $\bar{\rho}$, and the material's yield strength σ_y , where the value of σ_y is determined by $\dot{\sigma}_A$ (Age hardening rate), $\dot{\sigma}_{ss}$ (Solute hardening rate), and $\dot{\sigma}_{dis}$ (Dislocation hardening rate). The precipitate nucleation and growth rate $\dot{\bar{r}}$ are related to the dislocation density $\dot{\bar{\rho}}$. The evolution of yield strength with ageing time based on Zhan's model is presented in Figure 2.9.

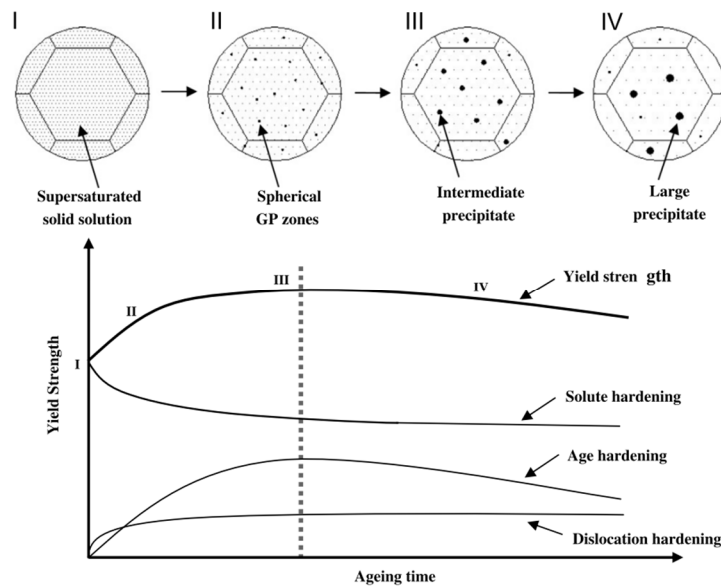


Figure 2.9. Age hardening mechanism of aluminium alloys based on Zhan's model [48].

Zhan's equations were originally developed for modelling the age hardening effect of aluminium alloy AA7055 from the solution heat treated condition to the over-aged condition. This set of equations was applied to aluminium alloy AA2219 by Yang et al. [50, 159], and to aluminium alloy 7B04 [160]. The equation was also implemented into ABAQUS for assessing the local curvature and springback variations [161] and coupled with surface residual stress [162] for formed parts in creep age forming.

In Chapter 5, this set of equations will be expanded so as to enable the modelling of the age softening effect in the peak aged condition for the AA7050-TAF alloy. In order to model the anelastic creep strain recovery effect after unloading, a new set of equations will be developed for modelling the inelastic/anelastic behaviour of the material.

2.5 Summary

The role of microstructural evolution in a material's strength, creep deformation, and anelasticity has been reviewed in this chapter. Whereas the theories and mechanisms behind anelasticity of materials after creep are quite well developed, how the creep age forming condition can affect anelasticity is still not fully understood. There is still a lack of knowledge about the relationship between anelasticity and springback. In terms of constitutive modelling, some models can already predict anelastic strain recovery; however, this behaviour has not been applied in modelling for creep age forming. The work in the upcoming chapters will focus on characterising anelasticity through mechanical testing so as to develop a constitutive model of anelasticity.

3 EXPERIMENTAL TECHNIQUES

3.1 General introduction

In this chapter, the experimental techniques for this study will be presented. Common mechanical testing methods including pure aging tests, uniaxial constant load creep aging tests, hardness tests, and tensile tests were used to examine the material's creep aging properties. Optical and transmission electron microscopy (TEM) were used to examine microstructural variations.

In the first part, methods of testing the creep aging properties of the AA7050-TAF alloy will be presented. Two steps of mechanical testing were designed. The first step was pure aging and uniaxial creep aging for different applied stress levels and aging time at the aging temperature of 174°C, and the second step was room temperature tensile tests for pure aged and creep aged specimens. Hardness tests were performed on pure aged and creep aged samples to evaluate the age hardening behaviour of the alloy. The experimental results from creep aging and room temperature tensile tests were used for determining the material constants of Zhan's model, which will be discussed in Chapter 5.

For investigation of the cause of anelasticity and its possible influence on springback, one important step is to conduct experiments to evaluate the mechanical behaviour of the material in use and to identify the mechanism of the creep age forming process on a microstructural scale. Experimental techniques for evaluating anelasticity and internal stresses have been reviewed in the previous chapter. For this study, new uniaxial

experimental methods, namely, the creep aging and recovery test (CART) and the creep aging and back stress test (CABT), were developed to evaluate the material's springback response after unloading from creep aging. It should be noted that the CART is a modified form of the strain relaxation and recovery test (SRRT) proposed by Sinha [4]. The CABT is a new technique developed by the author.

The CART and CABT are both designed to be performed under uniaxial conditions. In order to perform tests that can reproduce the actual forming conditions, where the stress distribution on the forming parts is more complicated than that in uniaxial loading condition – i.e., formed parts subjected to compression stress in the top and tension stress in the bottom of the plate – an *in situ* three-point bending stress relaxation and back stress test (SRBT) was conducted on the AA7050-TAF alloy. Table 3.1 lists the major difference and key control variables for CART, CABT and SRBT.

Table 3.1. Key control variables and their values for CART, CABT and SRBT.

	Time < 8h	Time > 8h
Creep aging and recovery test (CART)	137.5, 150, 162.5 MPa (Constant stress)	Stress free
Creep aging and back stress test (CABT)	137.5, 150, 162.5 MPa (Constant stress)	Unload to stress free condition then hold strain as constant
Three-point bending Stress relaxation and recovery test (SRBT)	From 2mm to 7.88mm (Constant deflection)	Unload to stress free condition then hold strain as constant

3.2 Materials and Specimen design

The material used in this study, aluminium alloy AA7050-TAF, has extraordinary mechanical properties and high fracture toughness. The basic specifications of this alloy follow the standard AMS 4050 [163], and the normal composition is shown in Table 3.2.

Table 3.2. Chemical composition of aluminium alloy AA7050-TAF (wt%).

Al	Zn	Mg	Cu	Fe	Si	Other
Bal.	6.25	2.14	2.23	0.05	0.03	0.15

The heat treatment process history of this alloy is based on AMS 2770 [164] with minor modifications in aging time in the second step aging. The reason for this is to obtain an

intermediate temper condition for age forming, namely, the Temper to Age Forming (TAF) condition. The heat treatment process of this material is illustrated in Figure 3.1. The material was first solution heat-treated at 477°C for 1h and then water quenched. It was then subjected to first-step-aging at 121°C for 8h and further second-step-aging at 165°C for 1h. For material in this condition, it takes another 8h of aging time at 174°C to reach the equivalent state of T74 temper [165].

The geometry and dimensions of the specimens for the uniaxial creep aging test, creep aging and recovery test (CART), and creep aging and back stress test (CABT) complied with the ASTM E8 standard as shown in Figure 3.2. Test specimens were machined in the rolling direction from the as-received temper state.

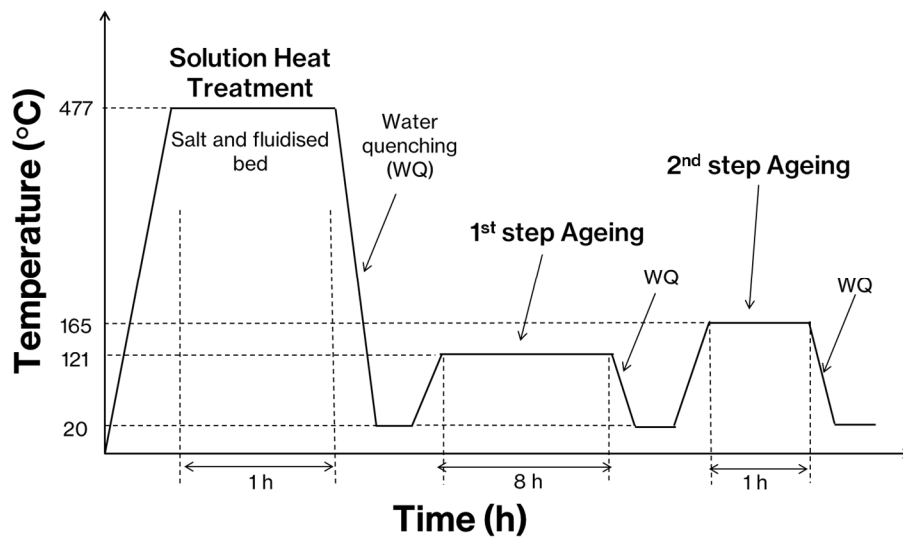


Figure 3.1. Heat treatment history of the as-received aluminium alloy 7050-TAF.

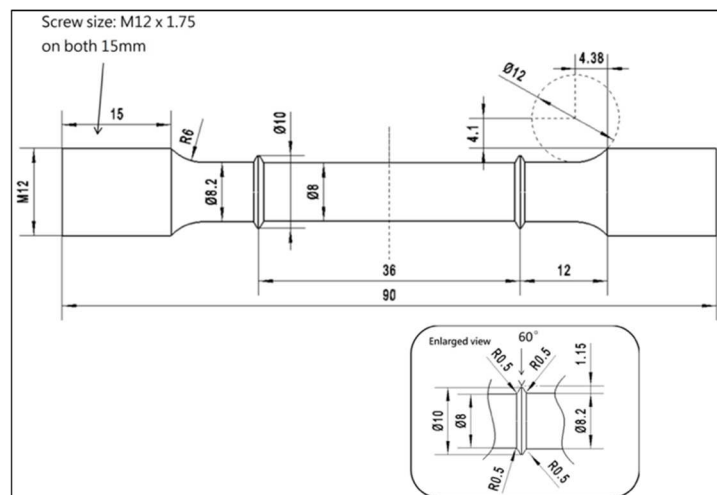


Figure 3.2. Dimensions of the specimens for uniaxial creep aging tests and subsequent tensile tests, CART, and CABT.

3.3 Experimental facilities and tools

In this section, the experimental facilities and details of the experimental procedure for pure aging, uniaxial creep aging tests with the subsequent hardness tests and tensile tests, creep aging and recovery tests (CART), creep aging and back stress tests (CABT), and the three-point bending stress relaxation and back stress test (SRBT) are explained, including the tools, instruments, and the methods used in these tests.

3.3.1 Machines and tools for pure aging and creep aging

The pure aging test was performed in an air circulated oven made by Lenton, as shown in Figure 3.3a, and the uniaxial constant load creep aging test was performed in a creep machine made by Phoenix Instrument, as shown in Figure 3.3b.

The creep machine is designed for tension creep tests at elevated temperatures. The machine contains a fulcrum loading system, a lower grip assembly, and a furnace. A load can be applied on the test sample by declining the lower grip assembly to lift the weight through the fulcrum arm. The load ratio of the fulcrum loading system is 10:1.

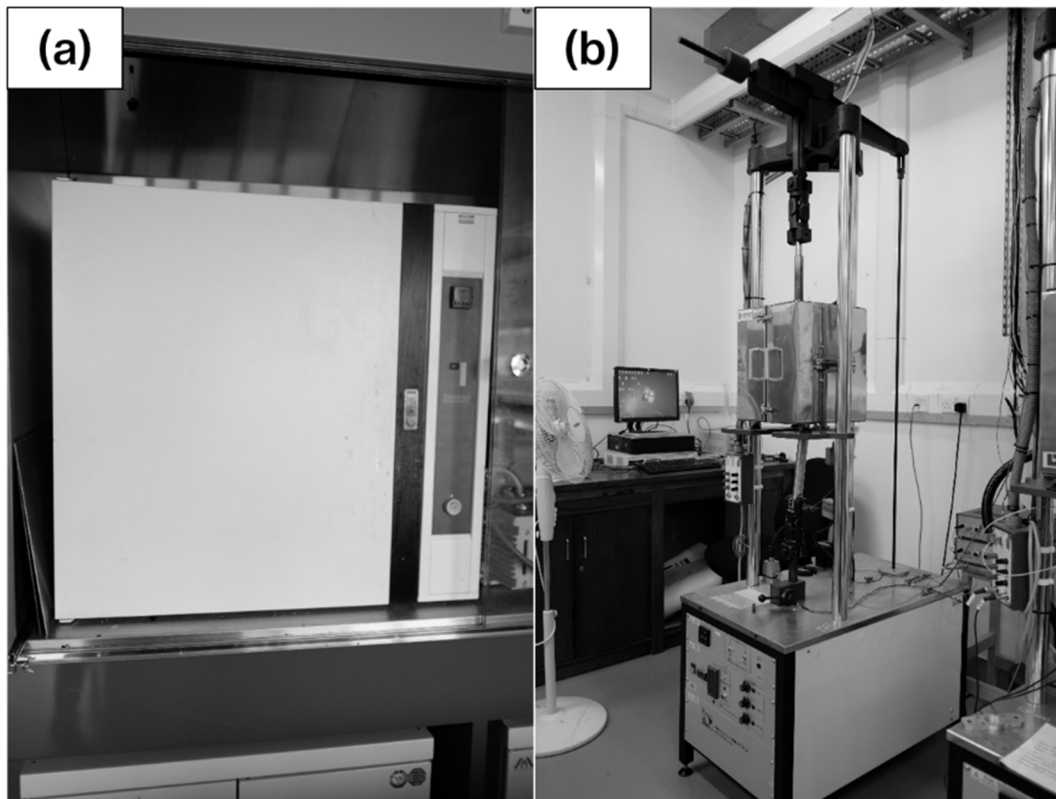


Figure 3.3. Testing machines for (a) pure aging, and (b) creep aging.

Figure 3.4 shows the tools and instruments required for uniaxial creep aging tests. For strain measurement during the creep aging test, single-electrode capacitance gauges were applied on the specimen. Capacitance gauges are direct metrology devices providing non-contact measurements of the displacement of the sample. They use an electric field to measure changes in capacitance between the probe and a conductive target surface, without physical contact. Two gauges were paired for this test. The gauges can detect changes in voltage as small as 1mV, which is about 0.5 μ m if converted into displacement.

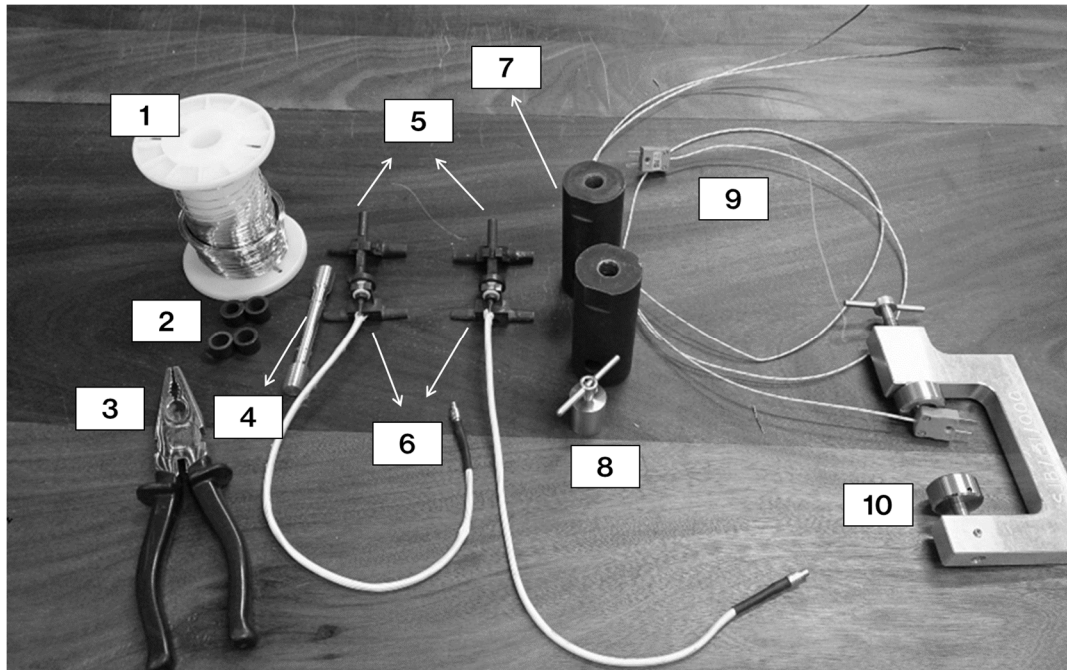


Figure 3.4. The tools and instruments for the uniaxial creep aging test (1. Wire; 2. Clamps for the capacitance gauges; 3. Pliers; 4. Test samples; 5. Upper parts of capacitance gauges; 6. Lower parts of capacitance gauges; 7. Rigs; 8. Clamp remover; 9. Thermal couples; 10. C-Clamp).

A capacitance to voltage converter, shown in Figure 3.5, is required to obtain the voltage value on the capacitance gauges. It is a device that can transduce the distance between upper and lower capacitance gauges into a voltage reading.

To transform a voltage value into a displacement value, it is necessary to know the relation between the voltage variations and the displacements. The voltage readings on the converter and the displacement between the upper and lower parts of the capacitance gauge follow a linear relationship. An INSTRON High Magnification Extensometer Calibrator (IHMEC), shown in Figure 3.6, was used to determine this relationship. The IHMEC calibrator consists of a micrometre head mounted on a stand frame, with two couplings holding various calibrator spindles. It enables the gauge to move over its

displacement range and permits the accurate adjustment of a gauge over very small distances for tensile calibration.

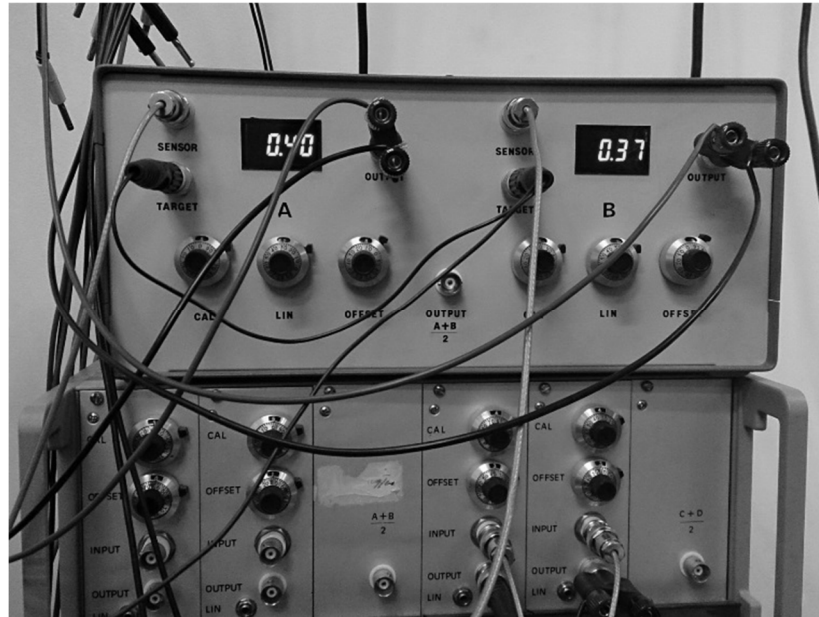


Figure 3.5. Capacitance to voltage converter.

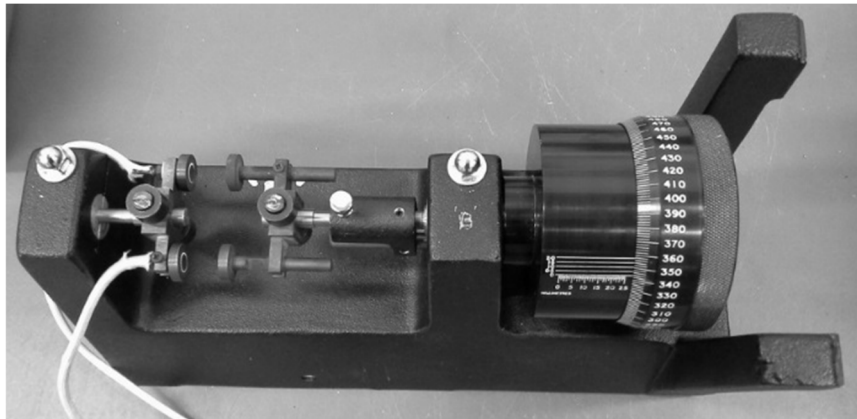


Figure 3.6. INSTRON High Magnification Extensometer Calibrator (IHMEC).

3.3.2 Machines and tools for tensile tests, CART, CABT, and SRBT

A computer-controlled INSTRON 5584 testing machine (as shown in Figure 7) equipped with an electromechanical load frame and a 150KN load cell was used for the tensile tests, CART, CABT, and SRBT. The accuracy of the load offered by the machine is 20 to 40N. A high temperature environment was provided by an assisting oven, as shown in Figure 3.7. The furnace can limit the temperature to a range of 2°C during the tests.

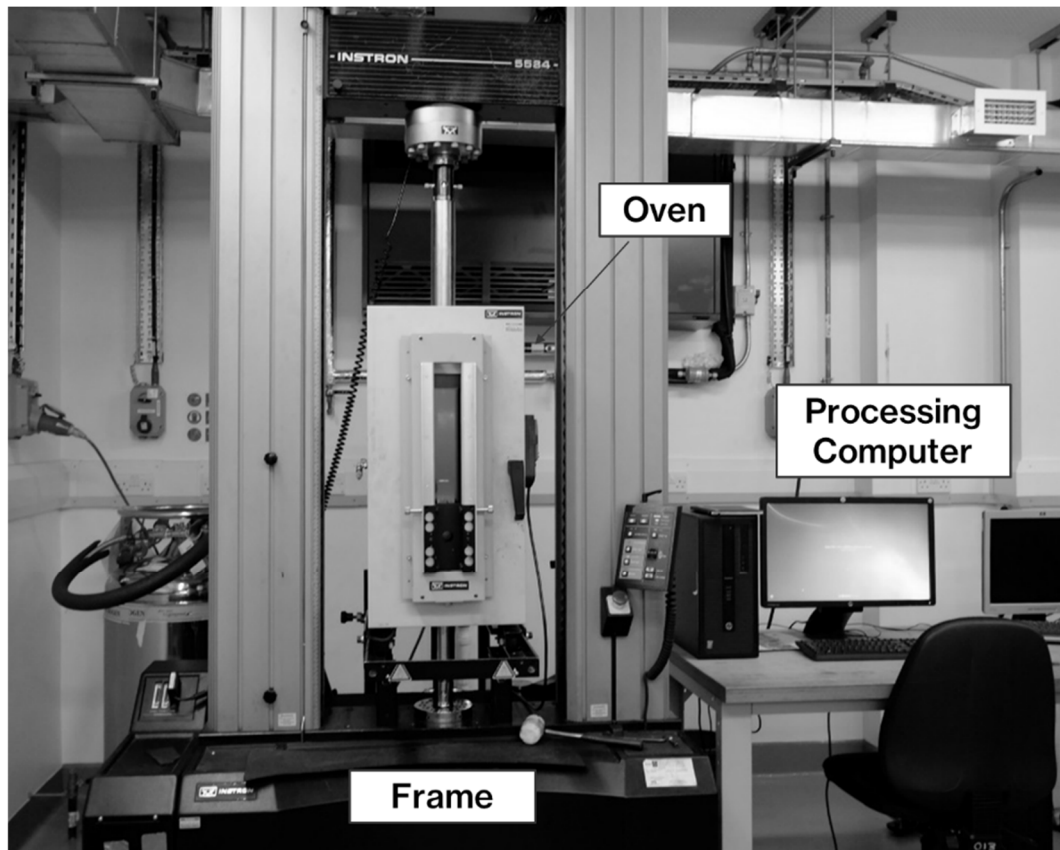


Figure 3.7. INSTRON machine 5584 with oven.

Figure 3.8 shows the tools and instruments required for these tests. A long axial travel distance extensometer, the Instron 2630-107, was used to measure the strain and whole elongation of the material in the tensile tests. A dynamic extensometer, the Instron 2620-601, was used in the CART and CABT for highly accurate measurement at the creep aging temperature.

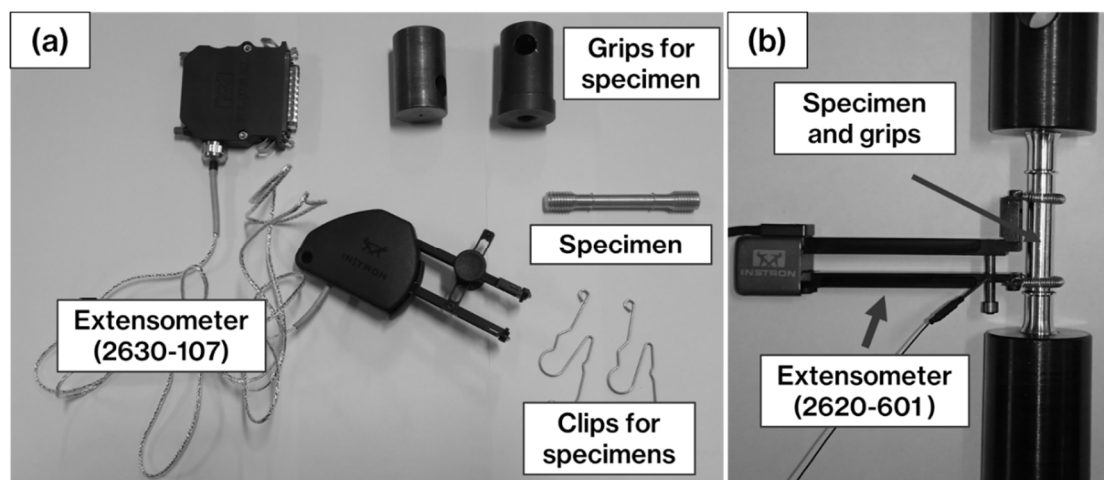


Figure 3.8. (a) An Instron 2630-107 extensometer for tensile tests. (b) An Instron 2620-601 extensometer for CART and CABT.

3.4 Experimental setup and procedures

This section delineates the setup and procedures for the experiments performed in this thesis. A schematic plot of the different tests is illustrated in Figure 3.9.

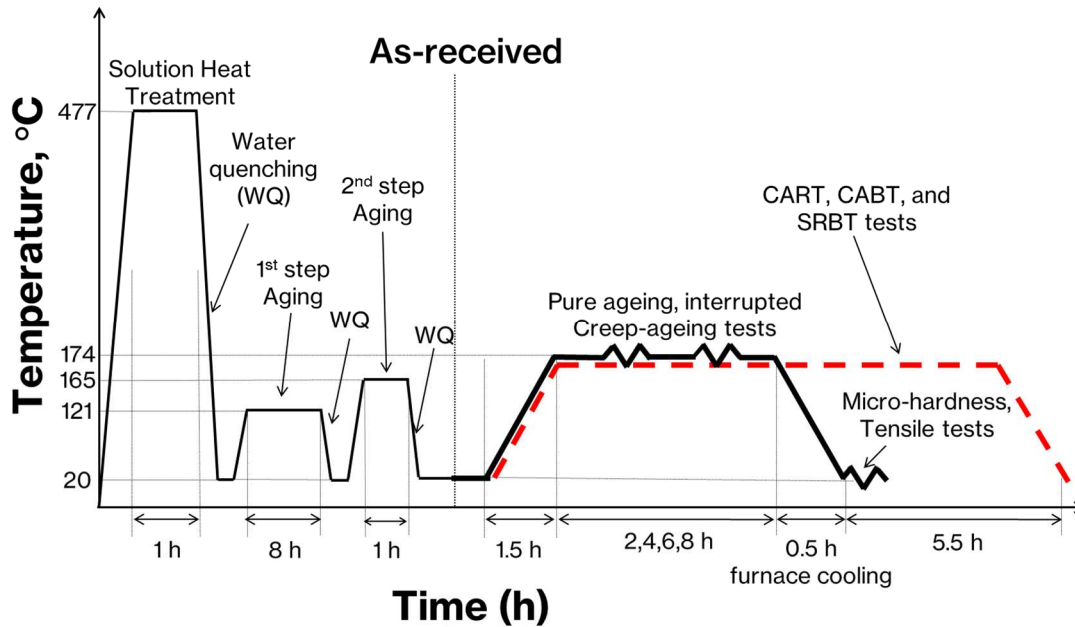


Figure 3.9. Heat treatment process and testing procedure.

3.4.1 Pure aging and creep aging treatments

The pure aging and uniaxial constant load creep aging tests were conducted in air at 174°C for a designated time. The times and temperature for these tests were slightly modified from AMS 2770 [164] so as to achieve the optimum performance of the material. The creep aging tests were conducted in air with eight different stress levels, 75, 100, 112.5, 125, 137.5, 150, 162.5, and 175 MPa, at the intermediate temperature range of 174°C for 2, 4, 6, and 8h. The creep strain was measured using two linear capacitance gauges fitted on the ridges of the samples to measure the displacement.

3.4.2 Hardness tests

Vickers Hardness tests were carried out using a Zwick ZHU0.2 universal hardness testing machine on the as-received, pure aged, and creep aged samples to identify the precipitation hardening behaviour. The testing method followed ASTM E384 standards [166]. The value of hardness was measured by forcing the Vickers indenter into the surface of the material at a load of 50 N and an average rate of 30 $\mu\text{m/s}$. The load application time for

one indentation was about 15 s, and the average hardness value for each specimen was determined from 25 tests on the corresponding sample.

3.4.3 Tensile tests

Tensile tests for the as-received and creep aged specimens were conducted on the Instron 5584 at a strain rate of 10^{-4} s^{-1} at room temperature and elevated temperature (174°C), respectively. Similar to that of performing hardness tests, the purpose of performing tensile tests is to determine the evolution of yield strength with creep aging time and the applied stress in creep aging.

The specimens were first locked onto the rods and the machine. The extensometer was then clamped on the specimen for the measurement of strain in the tests, as shown in Figure 3.10.

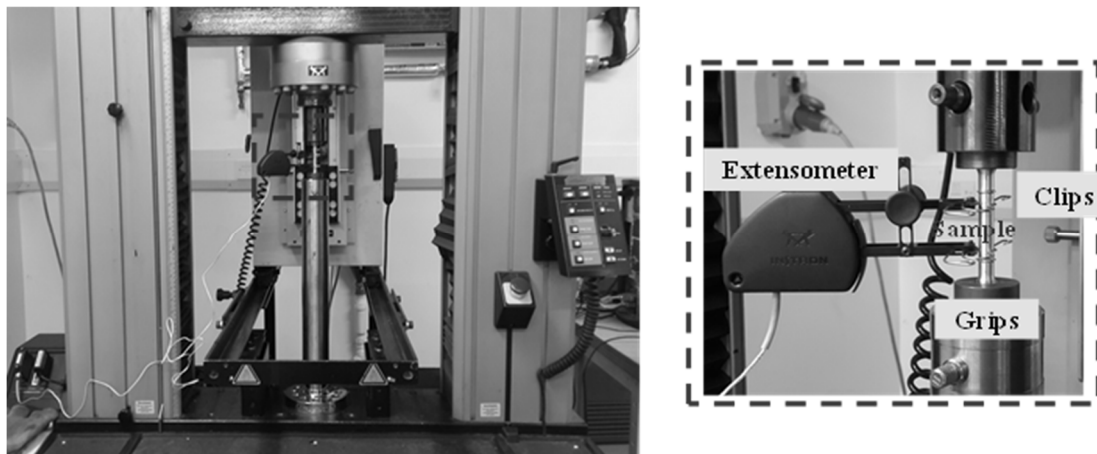


Figure 3.10. Configuration for the tensile test in the INSTRON 5584.

3.4.4 Creep aging and recovery test (CART)

The creep aging and recovery test (CART) is an extension to the uniaxial constant load creep aging test. The testing technique was modified from the Strain Relaxation and Recovery test (SRRT) developed by Sinha [4]. Unlike constant load creep aging tests, which measure only the creep strain generated during creep, the CART focuses on the inelastic strain and anelasticity after unloading from creep aging. A detailed introduction and explanation of the CART will be provided in Chapter 4.

In this CART series, the tensile load was applied at a rate of 8 MPa/s; such speed loading was adopted to minimise the inelastic strain occurring with time. After application of the

load, the specimens were held at constant stresses of 137.5, 150, and 162.5 MPa for 2, 4, 6, and 8h respectively to facilitate examination of the influence of creep aging time on anelasticity. At the end of the creep aging period, the tensile load was removed rapidly at a rate of 50MPa/s, and the strain continued to be monitored under the stress-free condition until the end of the strain recovery. Figure 3.11 shows the (a) creep aging and recovery curve and (b) loading history for the CART.

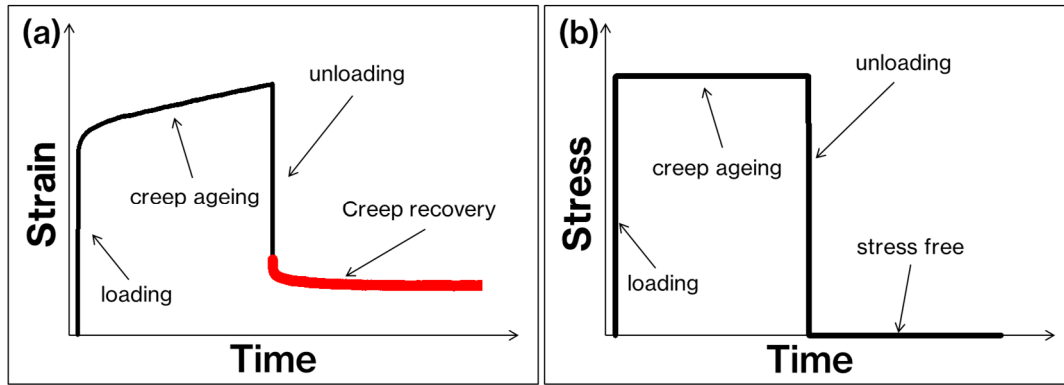


Figure 3.11. (a) Creep aging and recovery curve. (b) Loading history of the CART.

3.4.5 Creep aging and back stress test (CABT)

The creep aging and back stress test (CABT) is a new testing technique developed by the author. The idea of developing this new test arose from the need to quantify the internal stresses that cause anelasticity in the CART. A full description of the CABT will be provided in Chapter 4.

Instron 5584 testing machine introduced in section 3.3.2 was used for this test. Coupled with the Instron 2630-107 extensometer, the test not only provides linear loading and unloading rates but also captures the strain variations during creep aging and stress evolution after unloading, which are essential in performing this test.

Figure 3.12 illustrates the testing programme of the CABT. The tests were performed by applying a constant stress (137.5, 150, 162.5 MPa) to a specimen for 5, 10, and 30 min and for 1, 2, 4, 6, and 8h, just like the conventional creep aging test. The specimens were then unloaded to the stress-free condition, where the strain was upheld as a constant, as shown in Figure 3.12a. The stress variation with time (back stress) was also captured, as shown in Figure 3.12b.

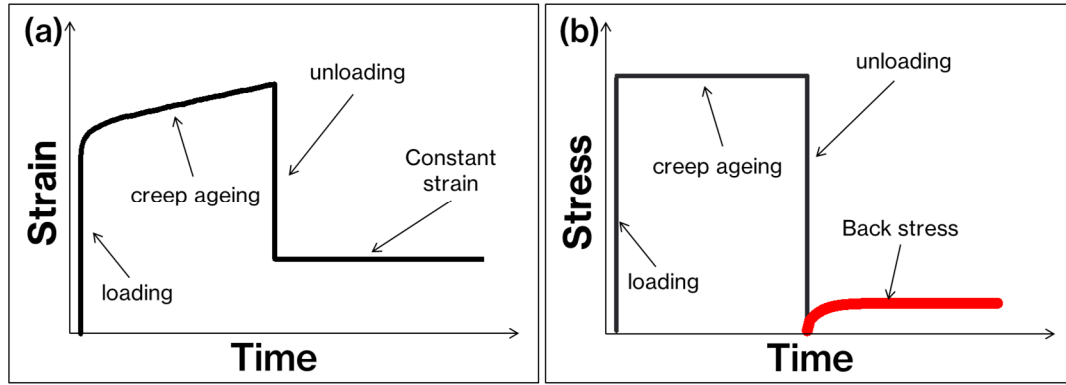


Figure 3.12. Illustration of CABT: (a) A creep ageing total strain-time curve where the strain was held constant after unloading. (b) Loading history of CABT where the back stress was generated after unloading.

3.4.6 Three-point bending stress relaxation and back stress test (SRBT)

Constant strain three-point bending SRBTs for three thicknesses of material, 3mm, 5mm, and 8mm with the width of 25mm, were carried out at 174°C. An Instron 5584 testing machine with a temperature chamber was used for the tests, as shown in Figure 3.13. A three-point bending rig was placed in the temperature chamber so that the rig could be directly controlled by the load frame. The load cell on this machine provides a load weighing system for measuring load changes during stress relaxation and back stress after unloading. This method overcomes some past difficulties in acquiring experimental data in the creep ageing forming process, such as the direct measurement of the stress relaxation rate during forming and springback, and the effects of stress on springback.

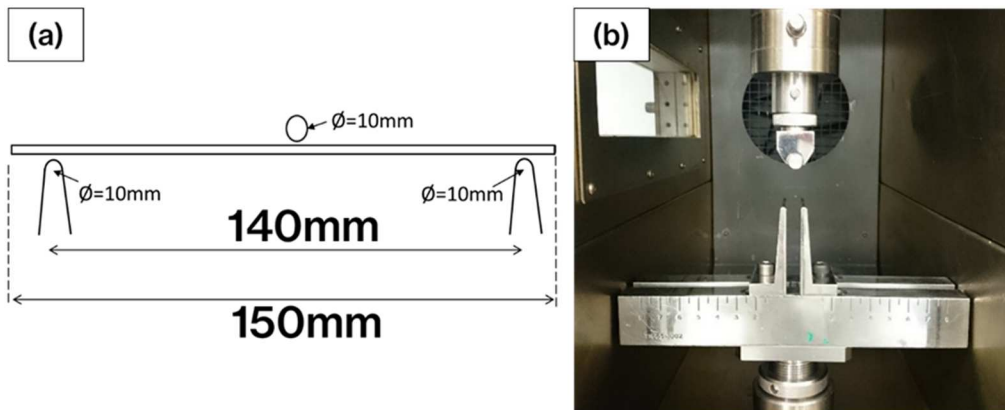


Figure 3.13. (a) Dimensions and configuration of SRBT test. (b) A three-point bending rig installed in the temperature chamber.

The load was applied to the plate at a rate of 0.2mm/s, and the sample was held constant for 8h for the stress relaxation process. The load was then removed at the same rate. When the load reached 0N, the strain was held constant to monitor the load changes at this stage.

This stage was carried out to evaluate the internal stress induced at this stage. Figure 3.14 illustrates the displacement and loading configuration of the SRBT.

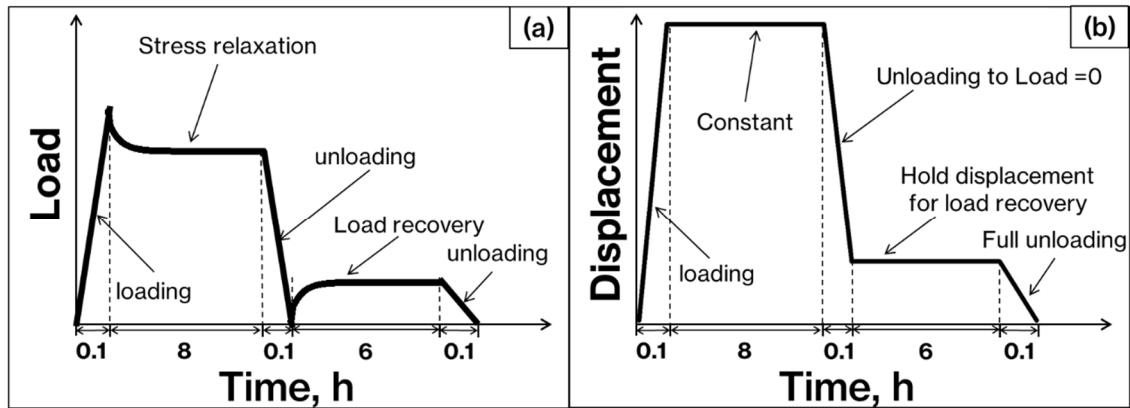


Figure 3.14. Schematic diagram of the loading and displacement configuration for the three-point bending stress relaxation and back stress test (SRBT).

3.5 Microstructural Characterisation

3.5.1 Optical Metallography

The shape, size, and morphology of the grains of the as-received AA7050 alloy were examined using optical metallography. The samples were mounted in a resin block, ground with abrasive papers, and then polished with alumina powder to produce a scratch-free finish. The polished specimens were then etched for 3 to 5 min with Keller's reagent, consisting of 95ml water, 2.5ml HNO_3 , 1.5ml HCl , and 1.0ml HF , by immersing the specimen in the etchant. After the etching, the specimens were examined with an optical microscope, a Zeiss Axio Scope.

3.5.2 Transmission electron microscopy (TEM)

Selected as-received, pure aged, and creep aged specimens were examined with transmission electron microscopy. Since the material (AA7050-TAF) is designed to reach T74 condition after 8h of aging, TEM was only performed on the samples of up to 8h pure aging and creep aging. Two samples after CART and CABT were selected for identifying dislocation cell structures, which will be presented in section 4.3.4. The TEM specimens were cut from gage sections along the applied stress directions. They were first cut into 3mm diameter discs and then ground mechanically to 0.08mm thickness. The thin foil specimens were then electro-polished with a twin-jet electropolisher to perforation in a

solution composed of 1/3 nitric acid and 2/3 methanol at -20°C and a voltage of 10V. The electro-polished specimens were then examined in a Tecnai F20 (200 kV) field-emission-gun scanning transmission electron microscope equipped with an energy dispersive X-ray (EDX) spectrometer.

3.6 Summary

Conventional pure aging and creep aging tests were carried out to investigate the creep aging properties of AA7050-TAF alloy. Newly-developed tests, the CART and CABT, were conducted to characterised the internal stresses, namely, the back stress generated during the forming process of the material, and to evaluate the influence of internal stresses on anelastic creep recovery. The three-point bending SRBT was developed to reproduce the creep age forming process. The improvement in testing techniques, the use of load cells coupled with a computer controlled extension method, allowed the direct measurement of stress evolution in the forming process. Microstructural characterisation by TEM revealed the nanometer-sized precipitates in aluminium matrix. Full experimental results will be presented in the next chapter.

4 EXPERIMENTAL RESULTS AND DISCUSSION

In this chapter, experimental results on the creep aging, microstructural evolution, anelasticity, and creep age forming behaviour of AA7050-TAF alloy will be presented and discussed with the aim of characterising the creep aging and anelastic properties of AA7050-TAF alloy so as to identify the deformation mechanisms that occur during the creep age forming process for the development of a creep age forming model that takes into account anelastic creep recovery.

4.1 Creep aging properties

4.1.1 Hardness of pure aged samples

The hardnesses of the as-received and the heat treated pure aged samples are shown in Figure 4.1. The hardness of the as-received alloy, 208HV5, is higher than those of the pure aged samples, which decrease with increases in aging time. This result indicates that the as-received alloy has already reached the hardness peak and possesses the optimum precipitate size and distribution in the matrix, which together contribute to the hardness. During pure aging treatment, the precipitates coarsen and the hardness consequently declines.

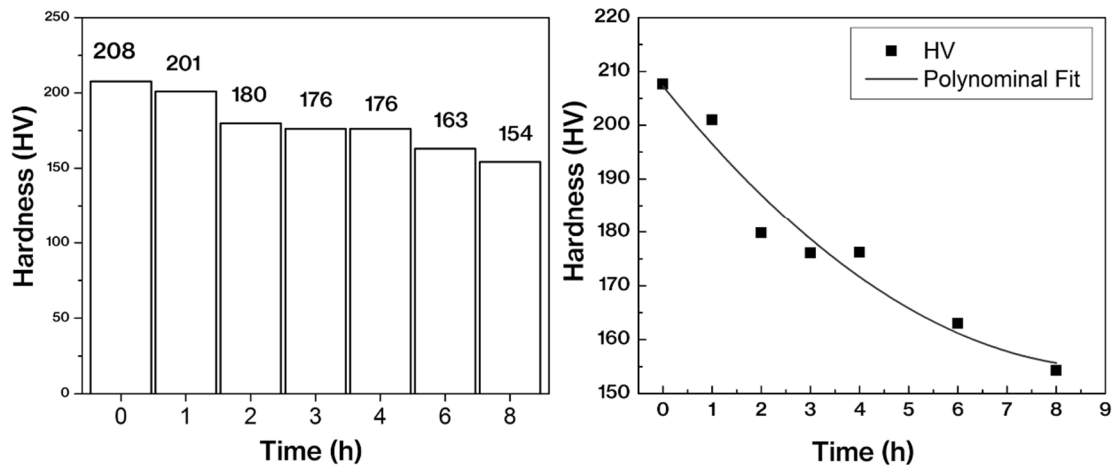


Figure 4.1. Hardness evolution of pure aged samples treated at 174°C.

4.1.2 Creep behaviour

Figure 4.2 shows the creep strain versus aging time curves at different stress levels. The primary creep stage generally occurs within 1h for all stress levels. The steady state of creep can be observed in this alloy under test conditions of below 162.5 MPa and less than 8h. At 175 MPa, however, tertiary creep occurs after 5h of testing.

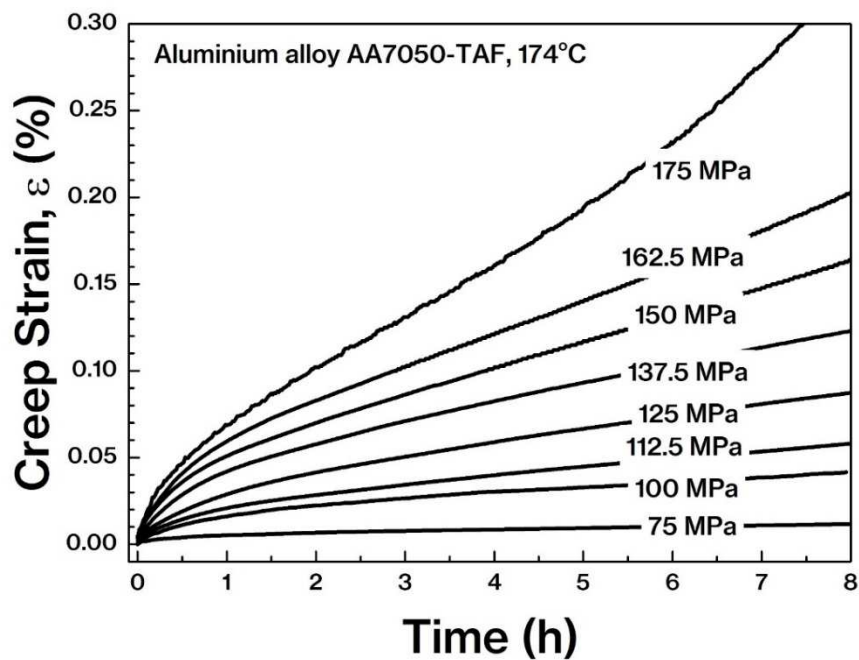


Figure 4.2. The creep strain versus time curves for different initial stresses.

4.1.3 Tensile properties

The strain rate for the tensile test was 10^{-4} s^{-1} at room temperature. The stress-strain curves of the as-received and pure aged specimens at 174°C are shown in Figure 4.3. The tensile properties of the as-received and pure aged AA7050-TAF alloy are listed in Table 4.1. It can be seen that flow stress decrease but the elongation increase with the increase in aging time, and that the as-received alloy has the highest yield strength, σ_y , and ultimate tensile strength, σ_{UTS} . The yield stress and the ultimate tensile stress decrease as aging time increases, which is consistent with the hardness results.

The elongation at break, on the other hand, first drops slightly from 18.6% (as-received) to 18.2% (2h of pure aging) and then increases with the aging time to 18.3% (4h of pure aging), 19% (6h of pure aging), and 20.5% (8h of pure aging). For the creep aged samples, however, the elongations at break are at 10-13%, which is much lower than those of the pure aged specimens. As shown in Table 4.2, this may be due to the micro-cracks generated on the surface of the sample during creep aging treatment.

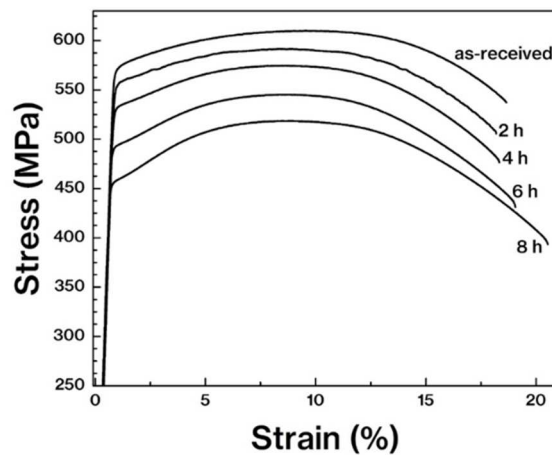


Figure 4.3. Representative stress-strain curves of aluminium alloy AA7050-TAF after different aging time treatments.

Table 4.1. Tensile properties of the as-received and pure aged samples.

Material	Aging time (h)	σ_y (MPa)	σ_{UTS} (MPa)	Elongation at break (%)	σ_{UTS}/σ_y
As-received	0	573	608	18.6	1.06
Pure aged	2	556	592	18.2	1.06
Pure aged	4	533	574	18.3	1.07
Pure aged	6	492	545	19.0	1.10
Pure aged	8	456	518	20.5	1.13

Table 4.2. Tensile properties of the creep aged samples under 150 and 162.5 MPa.

Applied Stress (MPa)	Aging time (h)	σ_y (MPa)	σ_{UTS} (MPa)	Elongation at break (%)	σ_{UTS}/σ_y
150	2	558	582	12.4	1.04
150	4	510	537	11.2	1.05
150	6	500	566	10.1	1.10
150	8	457	513	11.4	1.10
162.5	2	557	587	13.1	1.05
162.5	4	492	521	10.3	1.06
162.5	6	478	530	9.6	1.10
162.5	8	463	519	10.4	1.10

The yield strength (0.2% offset) data in Table 4.1 and Table 4.2 are plotted versus creep aging time, as shown in Figure 4.4. The material generally exhibits overaging behaviour; the yield strength decreases with increases in aging time. The maximum yield strength is about 573 MPa in the as-received condition. That value gradually drops to 533 MPa (after 4h) and 456 MPa (after 8h) under the pure aging condition.

For creep aged material, when the applied stress is increased to 162.5 MPa, the yield strength decreases significantly to 492 MPa (after 4h) and 463 MPa (after 8h). An increase in the applied stress during creep aging accelerates the decrease in the yield strength. As the creep aging time increases further to 8h, the yield strengths of the samples with the different applied stresses gradually approximate the same level.

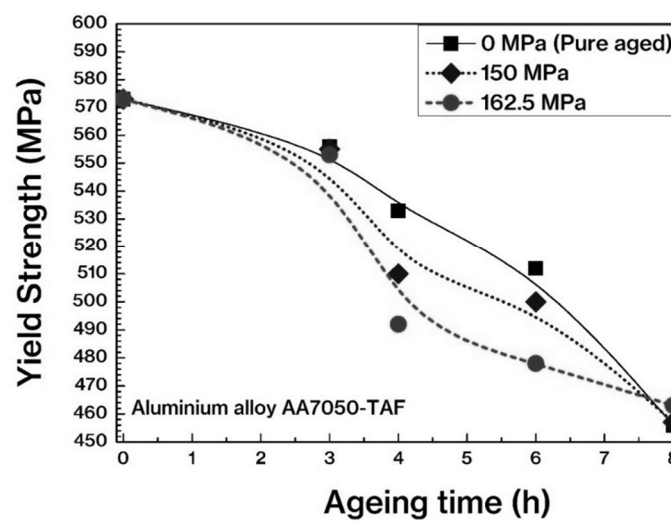


Figure 4.4. Yield strength of creep aged samples versus creep aging time.

4.1.4 Summary

The temper condition of the as-received AA7050-TAF alloy is close to the peak aged condition. An evaluation of the creep aging properties of this material at 174°C for up to 8h indicates typical overaging behaviour, as characterised by the time and applied stress dependence. Tensile tests on the pure aged samples show that elongations of the material increase as aging time increases. The result is consistent with increases in the ductility of the material during the aging process. However, the elongations are generally lower in the creep aged samples than in the pure aged ones. This may be due to the micro-cracks that occur during the creep aging process.

The curves for yield strength versus creep aging time indicate that the applied stress for creep aging can decrease the yield strength of the material; the higher the applied stress, the lower the yield strength. The difference in the yield strengths of the samples under different applied stresses, however, is minimised after 8h of aging.

4.2 Microstructural characterisation

The mechanical testing presented in the previous section produced an intriguing result: For the samples of AA7050-TAF alloy, their resulting yield strengths depend on not only the subsequent aging time but also the applied stress levels. A further microstructural investigation was performed to elucidate the micro-mechanisms for this phenomenon. The overall morphologies and grain shapes of the as-received material were first characterised by optical microscopy. The microstructural evolution of pure aging and creep aging was then characterised by transmission electron microscopy (TEM).

4.2.1 The as-received material

An optical metallograph (OM) of the as-received material taken from the section along the longitudinal rolling direction is shown in Figure 4.5. The as-received AA7050-TAF alloy exhibits elongated grain structures that are partially recrystallized with a mean size of 15 μm due to its manufacturing process of rolling and heat treatment.

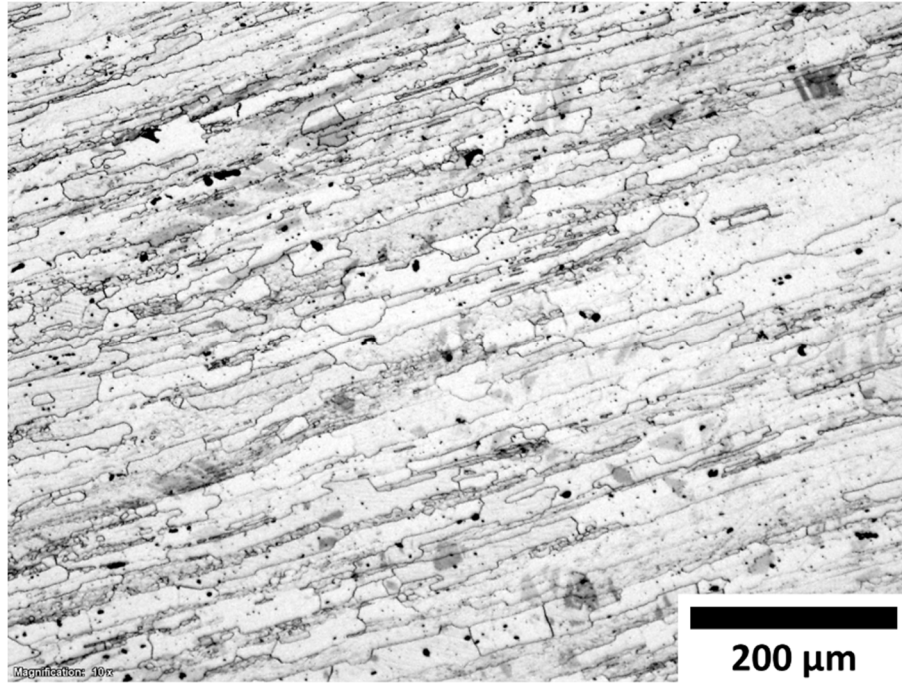


Figure 4.5. Optical metallograph of as-received aluminium alloy AA7050-TAF.

The montage of transmission electron microscopy (TEM) micrographs presented in Figure 4.6 shows that the OM elongated structures of as-received AA7050-TAF alloy actually consist of fine newly formed aluminium grains with a face centred cubic (fcc) lattice. The microstructure of as-received AA7050-TAF alloy contains GP zones and η' precipitates throughout the fine-grained matrix. These GP zones and η' precipitates are the dominant phases at this stage and contribute to the strength of the material [167]. Other larger precipitates and intermetallics, such as η phase (MgZn_2), $\text{Al}_7\text{Cu}_2\text{Fe}$, Mg_2Si , and Al_2CuMg , have also been reported in 7xxx alloy systems [168, 169].

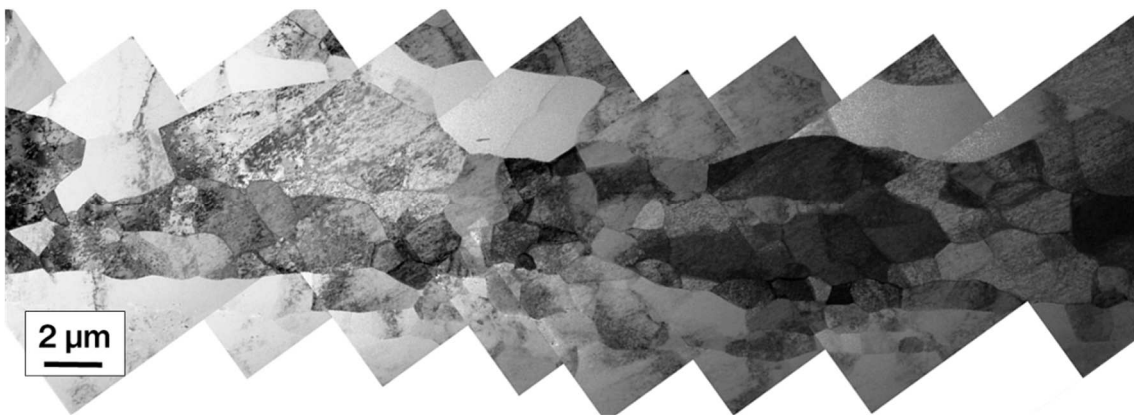


Figure 4.6. TEM micrograph showing the detailed morphology of grains.

Figure 4.7 presents the TEM micrographs of the as-received AA7050-TAF alloy. Tiny η' and η precipitates scattered within the Al matrix can be observed in Figure 4.7a; their size distribution is presented in Figure 4.7b. Very few dislocations can be seen at this stage. High resolution transmission electron microscopy (HR-TEM), as shown in Figure 4.7c, revealed GP zones (marked with 1-4), which are characterised by the several atomic-layer structure. Their corresponding fast Fourier transform (FFT) diffractograms are presented in Figure 4.7d; besides the main diffraction patterns of the aluminium fcc matrix, the characteristic diffuse streak pattern can be observed. The streak pattern originates from the atomic layer structured GP zone. It should be noted that the direction of streaking is perpendicular to the layers of the GP zone; i.e., the typical layer-structure's reciprocal lattice signal. An HR-TEM image of η' is presented in Figure 4.8a, and the corresponding FFT diffractogram, obtained from the moiré fringes of η' and the matrix, is shown in Figure 4.8b. The analysis of diffraction patterns is illustrated in Figure 4.8c. EDX data of Figure 4.8d indicate the precipitate η' to be rich in the elements Al, Zn, Cu, and Mg.

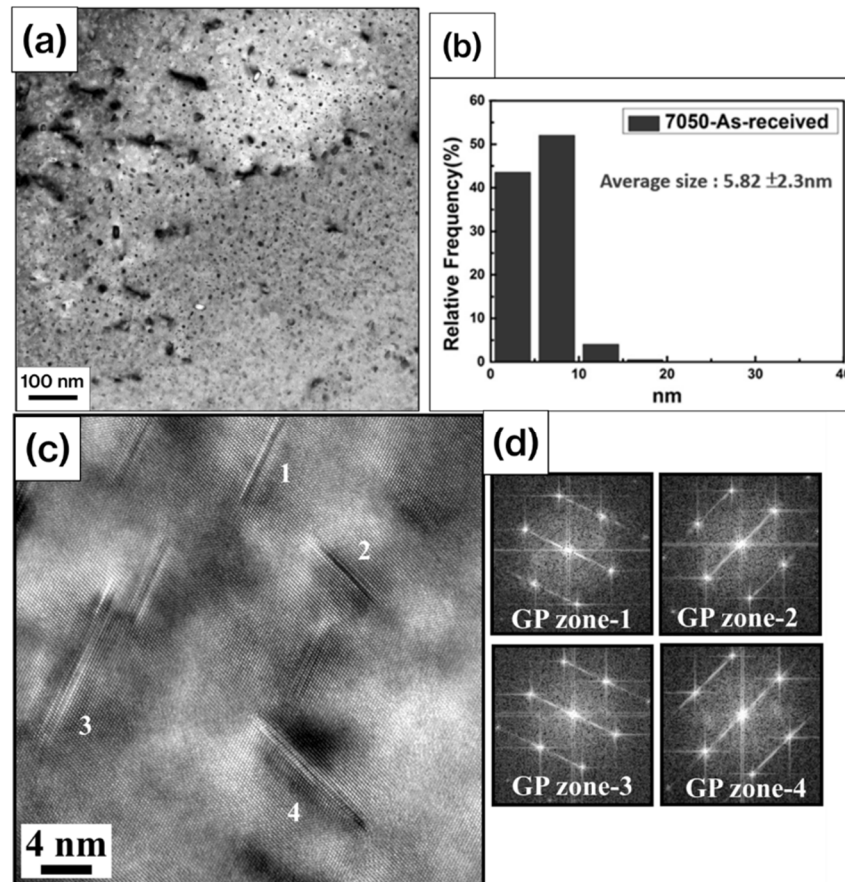


Figure 4.7. (a) TEM micrograph showing η' and η precipitates scattered within the Al matrix. (b) size distribution of the precipitates, (c) HR-TEM micrograph showing the GP zones (marked with 1-4), and (d) the corresponding FFT diffractograms of GP zones.

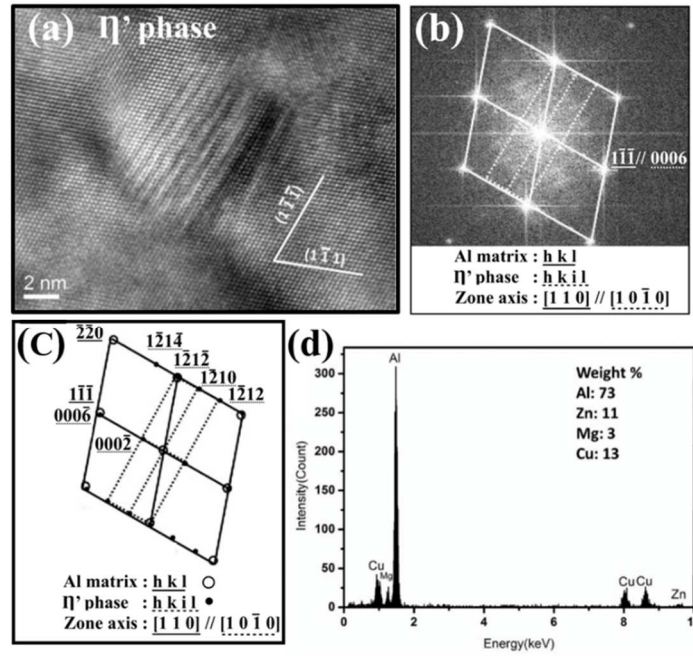


Figure 4.8. (a) HR TEM image showing the metastable precipitate η' in Al matrix; (b) corresponding FFT diffractogram obtained from (a); (c) schematic illustration of diffraction patterns for precipitates and Al matrix; and (d) corresponding EDX spectra of the precipitate η' .

4.2.2 Influence of stress on precipitation

For the investigation of the effects of stress and aging time on the microstructural evolutions, AA7050-TAF samples treated by pure aging and creep aging conditions were examined by TEM. During aging, GP zones easily transform into η' phases, thereby eliminating the strain field energy of the GP zones, which arises from the intrinsic rearrangement of Mg and Zn atoms. The η' phases possess four variants of orientation relationships with respect to the Al matrix: $(0001)_{\eta'} // (111)_{Al}$ and $[1\bar{1}00]_{\eta'} // [0\bar{1}1]_{Al}$; $(0001)_{\eta'} // (\bar{1}11)_{Al}$ and $[1\bar{1}00]_{\eta'} // [0\bar{1}1]_{Al}$; $(0001)_{\eta'} // (1\bar{1}1)_{Al}$ and $[1\bar{1}00]_{\eta'} // [011]_{Al}$; $(0001)_{\eta'} // (11\bar{1})_{Al}$ and $[1\bar{1}00]_{\eta'} // [011]_{Al}$. Therefore, under TEM observation along the $[1\bar{1}00]_{\eta'} // [0\bar{1}1]_{Al}$ zone axis, two variants of η' phases are in edge-on configuration along the $(\bar{1}11)_{Al}$ and $(111)_{Al}$ habit planes, appearing to be a rod-like morphology. However, the other two variants are not in edge-on configuration, appearing to be a disc-like morphology.

Figure 4.9 shows the TEM micrographs taken along the zone axis of $[0\bar{1}1]_{Al} // [1\bar{1}00]_{\eta'}$ for the pure aged and creep aged samples. The precipitates in pure aging and creep aging samples have very different morphologies. The direction of the applied tensile stress of

162.5 MPa for creep aging is indicated in Figure 4.9b, d, and f. The elastic stress had a strong impact on the creep aging samples, leading to the alignment of the precipitates perpendicular to the stress direction, as compared to the pure aging samples in Figure 4.9a, c, and e. TEM micrographs of the samples after 4h of pure aging (Figure 4.9c) and 4h of creep aging (Figure 4.9d) clearly show slightly larger precipitates than those in the samples after 2h of pure aging (Figure 4.9a) and creep aging (Figure 4.9b) samples. Samples after 8h of pure aging (Figure 4.9e) and after 8h of creep aging (Figure 4.9f) tend to possess coarser precipitates, which have been identified as η phase. It is appropriate to conclude that $\eta' \rightarrow \eta$ transformation occurs during the course of pure and creep aging treatments. However, the creep aging treatment causes the η precipitates to become elongated when TEM observation is along the zone axis $[0 \bar{1} 1]_{\text{Al}}$.

Figure 4.10 shows the TEM micrographs in a plane-view of the habit plane of the precipitates in pure aged and creep aged samples; i.e., the TEM sample was aligned under the zone axis parallel to $\langle 111 \rangle_{\text{Al}}$ so that the habit planes of precipitates (GP zones and η' precipitates) could be observed. Precipitates in plate, rod, and spheroid-like shapes can be seen. GP zones and η' precipitates can be distinguished through analysis of electron diffractograms. The morphologies in Figure 4.10 show that no distinct difference can be observed in shape and distribution of the GP zones and η' precipitates between pure aging and creep aging process.

In order to evaluate the differences in size of G.P. zones and η' precipitates obtained under different conditions, the measurements of the size of the GP zones and η' precipitates were conducted. As noted previously, GP zones and η' precipitates are the key contributors to age hardening, and the larger phases such as η precipitates have little effect on strength. The measurement on η precipitates was not carried out. In high-resolution TEM analysis, the crystal structures of the GP zones, η' precipitates and η precipitates can be identified by Fast Fourier transform diffractograms. Since the growth of GP zones and η' precipitates tends to expand their habit planes, the longitudinal lengths of them are employed to represent its overall size. The mean values of longitudinal length were determined measuring GP zones and η' precipitates from the TEM micrograph images. The average longitudinal length for each condition was determined from 50 measurements on the corresponding sample.

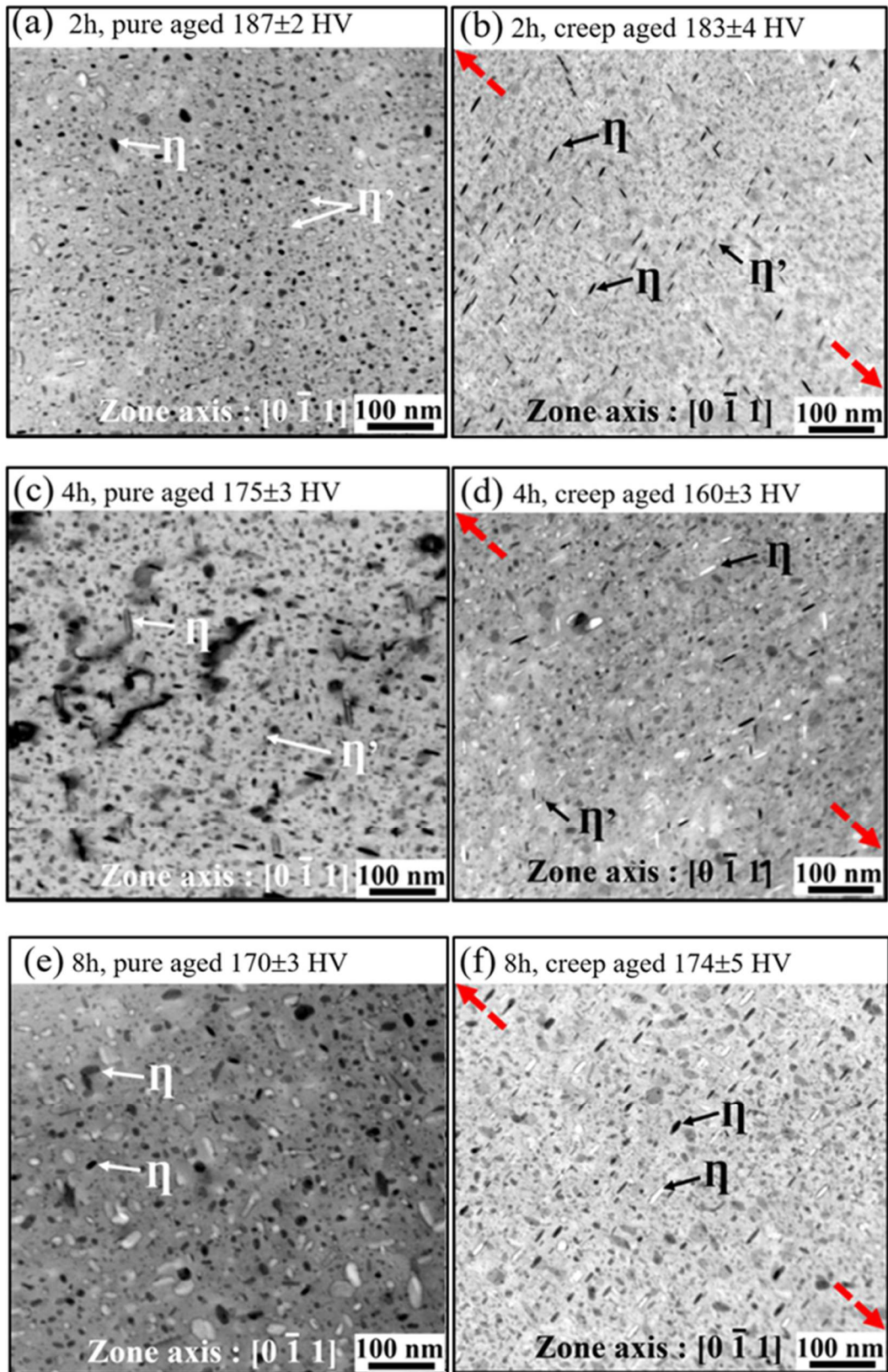


Figure 4.9. TEM micrograph of samples pure aged and creep aged at 174°C for (a)-(b) 2h, (c)-(d) 4h, (e)-(f) 8h (the direction of applied stress is indicated by red arrows).

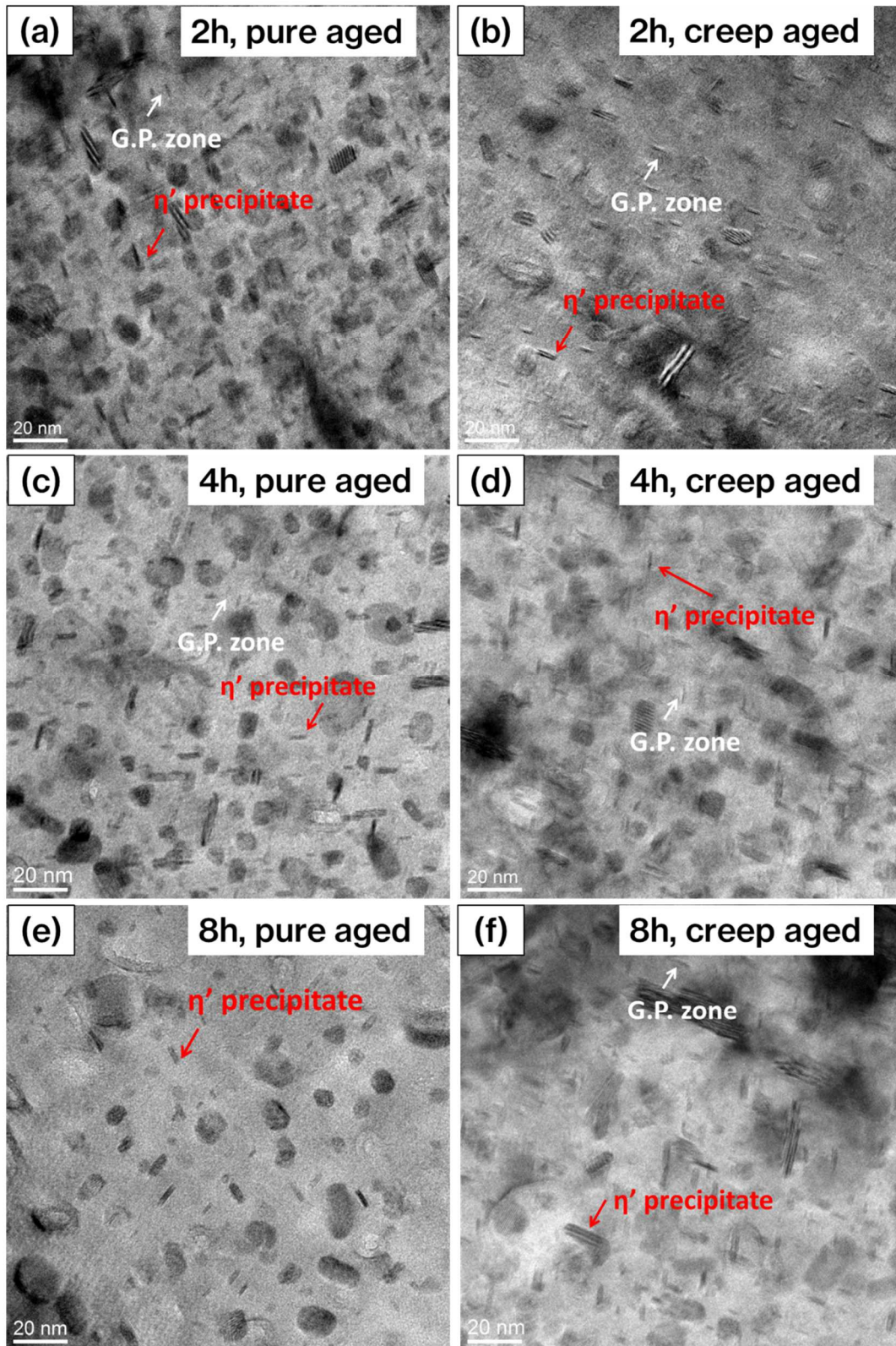


Figure 4.10. G.P. zones and η' precipitates of samples pure aged and creep aged for (a)-(b) 2h, (c)-(d) 4h, and (e)-(f) 8h.

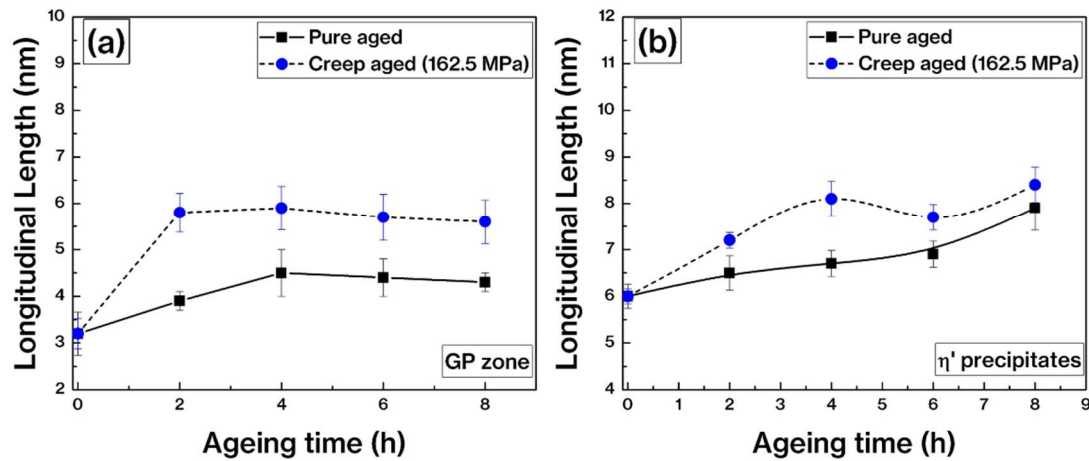


Figure 4.11. The evolution of GP zones and η' precipitates in longitudinal length with aging time for (a) GP zones and (b) η' precipitates under pure aged and creep aged conditions.

Figure 4.11 shows the measured values of longitudinal lengths of GP zones and η' precipitates under pure aging and creep aging conditions (with stress of 162.5MPa) as a function of aging time. For the pure aged sample, the longitudinal length of GP zones (Figure 4.11a) gradually increases from 3.2 nm in the as-received condition to 4.5 nm after 4h of aging and then stops growing. For the creep aged samples, on the other hand, the longitudinal length of GP zones (Figure 4.11a) increases more rapidly from 3.2 nm to about 6 nm and stops growing. The longitudinal lengths of GP zones in creep aged samples are generally 1.5 nm longer than those in the pure aged samples, as shown in Figure 4.11a. For the creep aged samples, the habit planes of GP zones are extended by the applied stress under the creep aging condition. Therefore, the longitudinal lengths of GP zones in the creep aged samples are longer than those in pure aged samples. It is presumed that the elongated GP zones gradually transform into η' precipitates. The latter are longer than the former. As shown in Figure 4.11b, η' precipitates are generally longer in the creep aging condition than in the pure aging condition. After 4h of aging, the length of η' precipitates reaches 8.1 nm for the creep aged samples, but only 6.7 nm for the pure aged samples. After 8h of aging, the difference of the length between pure aged samples and creep aged samples becomes smaller; i.e., 7.9 nm for the pure aged samples and 8.4 nm for the creep aged samples.

4.2.3 TEM characterisation of GP zones, η' precipitates, and PFZ zones

As GP zones and η' precipitates are nanometer-sized, they contribute most of the strength to age hardening. These two phases are the transition phases, which possess better crystallographic match with the aluminium matrix, forming a coherent or semi-coherent interface, whereas the final equilibrium precipitates such as η precipitates are incoherent with the aluminium matrix.

For the AA7050-TAF alloy in the as-received condition, where GP zones and η' precipitates are the dominant phases, the slight distortion between GP zones (η' precipitates) and the aluminium matrix provides coherency strain hardening, which makes it difficult for dislocations to cut through. As a result, the strength of the alloy at this stage is high. When the alloy reaches the overaging condition, these precipitates (the transition phases) transform into incoherent η precipitates, which are larger in size and smaller in quantity. These precipitates can no longer provide coherency strain hardening, but their strengthening mechanism is still dependent on dispersion hardening – the dislocations bypassing those precipitates.

Phase transformation of GP zones $\rightarrow \eta' \rightarrow \eta$ is a continuous process [170]. Figure 4.12 shows a HR-TEM micrograph of the GP zones after 6h of pure aging. The GP zones suddenly disappear after 5 seconds of observation. This disappearance illustrates that GP zones are very unstable and tend to dissolve to promote the transformation of η' precipitates during the aging process. The transition of η' precipitates to η precipitates, which are usually assumed to be MgZn_2 , was observed by HR-TEM, as shown in Figure 4.13. The HR-TEM (as shown in Figure 4.13a) clearly displays that η' and η were connected together. Their corresponding fast Fourier transform (FFT) diffractograms are presented in Figures 4.13b and c; the analyses of the diffraction patterns to verify η' and η phases are illustrated in Figures 4.13d and e. It is presumed that the transition from η' to η proceeded in due course. The nuclei of η , which formed on the interface between η' and aluminum matrix, gradually grew and replaced η' . Eventually η' disappeared, and η did not alter the basic η' morphology. It should be noted that the transition from η' to η involves the distribution of alloy elements; η is usually presumed to be MgZn_2 .

GP zones, on the other hand, can also be observed at all levels of aging time. It is therefore reasonable to presume that when the material reaches the overaging condition, enough solute atoms remain in the aluminium matrix to transform into GP zones. Therefore, coherency strain hardening still contributes strength to the material. The main cause of softening in overaging is the coarsening of η precipitates, which leads to a significant decrease in the quantity of GP zones and η' precipitates. Under the creep aging condition, the applied stress can accelerate the coarsening process of η precipitates.

In 7050 alloy, precipitate free zones (PFZs) also influence the mechanical properties of the material in the creep aging process. Grain boundaries act like efficient vacancy sinks for segregation of solute atoms, resulting in the formation of the stable precipitates along the grain boundaries. Figure 4.14 presents TEM micrographs of the PFZs in the samples after 8h of pure aging and creep aging, respectively. The widths of the PFZs in both cases are nearly the same. The effect of applied stress on PFZs seems to be not significant in the present observation.

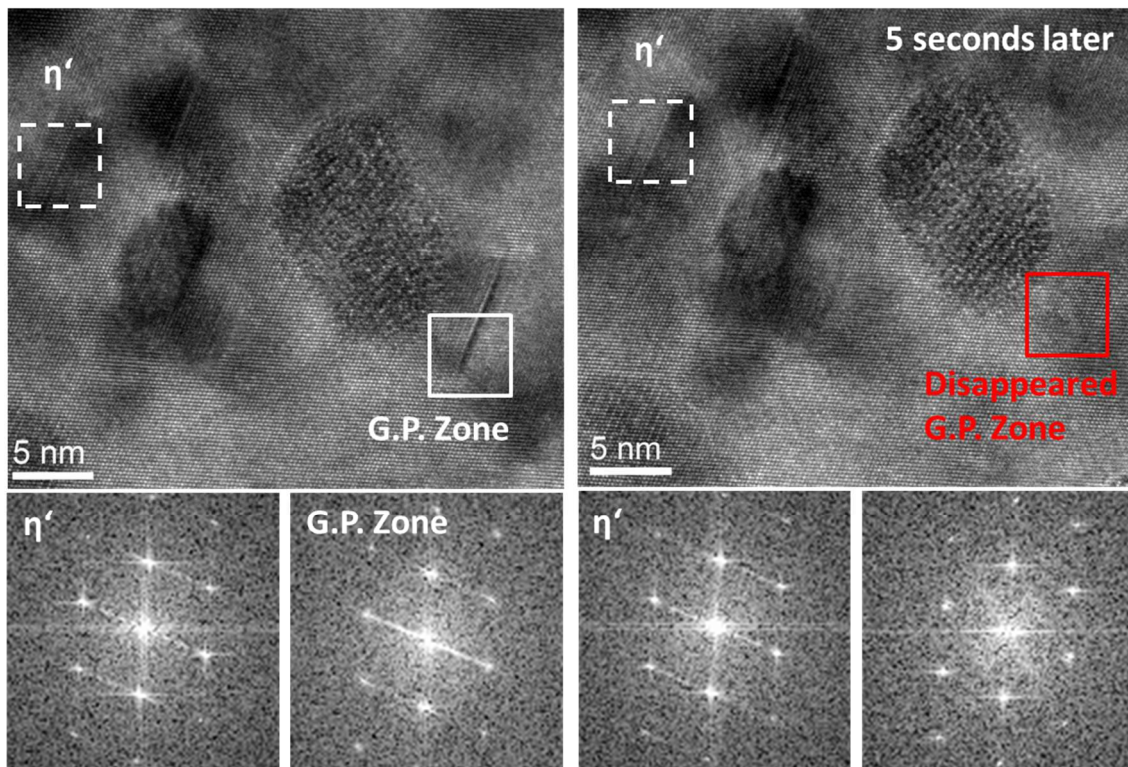


Figure 4.12. HR-TEM micrograph showing unstable GP zones in the sample with 6h of pure aging.

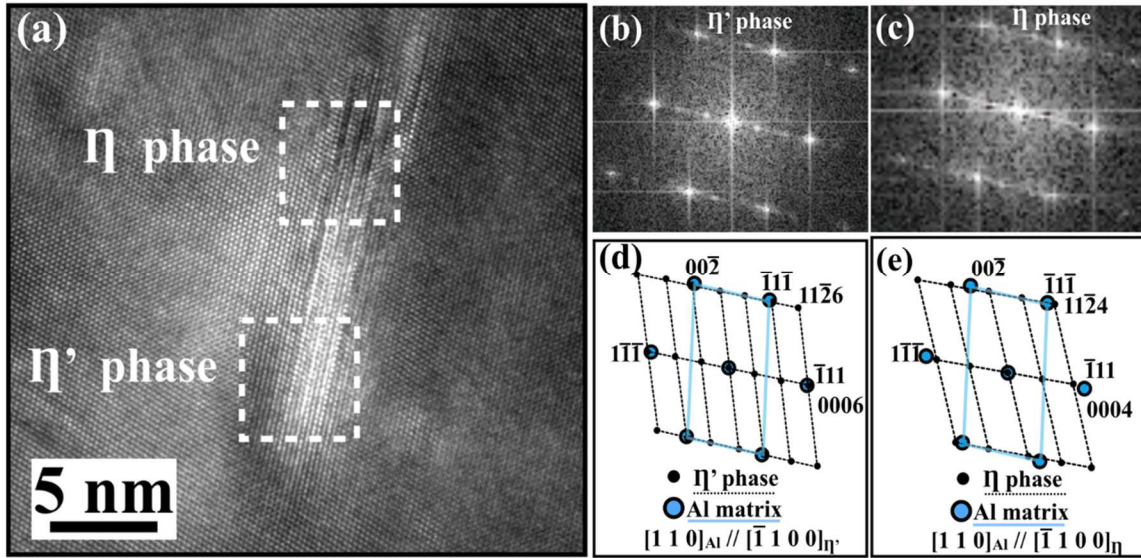


Figure 4.13. (a) TEM micrograph showing the transition from η' to η ; (b) and (c) corresponding fast Fourier transform (FFT) diffractograms; (d) and (e) analyses of the diffraction patterns.

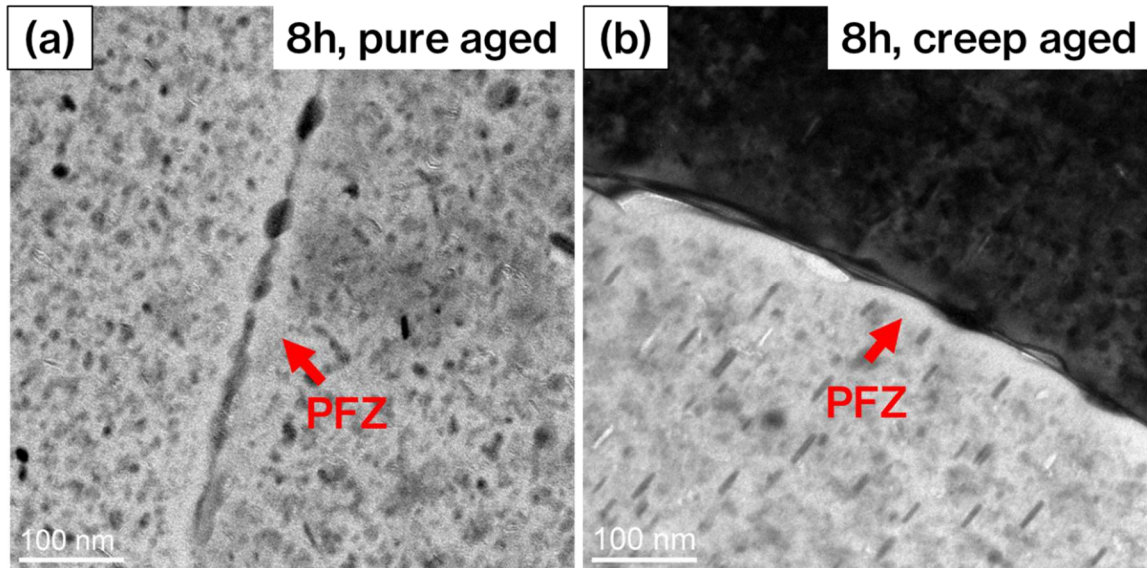


Figure 4.14. TEM micrographs showing the precipitate free zones (PFZs) of (a) a grain boundary of a pure aged sample; (b) a grain boundary of a creep aged sample.

4.2.4 Summary

TEM was used to characterise the precipitation behaviour under the pure aging and creep aging conditions. Current results show that the growth rate of GP zones and η' precipitates can be accelerated by applying stress, which also leads to faster coarsening of the η precipitates.

The results indicate that the precipitation of GP zones is a continuous process and even occurs after 8h of pure aging and creep aging, for GP zones can still be observed in the

aluminium matrix. The influence of the evolution of dislocation density on hardening could not be concluded, and further investigation on the measurement of dislocation density is required. Precipitate free zones can be found in both pure aging and creep aging samples.

4.3 Investigation of anelasticity with CART and CABT

The experimental study of the effect of anelasticity on springback in creep age forming is presented in this section. Two kinds of testing methods – the creep aging and recovery test (CART), and the creep aging and back stress test (CABT) – have been carried out on the AA7050-TAF alloy to investigate the significance of anelastic creep strain recovery and internal stresses in the springback and total deformation.

4.3.1 Introduction

An understanding of creep behaviour is essential not only for the development of creep resistant materials but also for building constitutive equations for creep age forming. Constant load creep aging tests are the conventional methods for this purpose. These types of tests mainly evaluate the relationship between creep strain, ε_c , minimum creep rate, $\dot{\varepsilon}_{min}$, and creep time under specified stress and temperature conditions. These tests are very useful for developing creep resistant materials and fracture life prediction because the main factors of interest are creep rate and time to rupture.

In the field of creep age forming, however, the main interests are the total amount of permanent deformation and the springback of a material generated during and after the forming process. Conventional constant load creep aging tests can only provide a measure of total creep strain – the sum of reversible and irreversible strains – but they do not provide information on each individual amount of time dependent reversible (anelastic) strain and irreversible (permanent) strain induced during forming.

As the formed parts will gradually recover from the deformation due to anelastic creep recovery, springback measured with conventional creep aging tests is insufficient to represent the entire deformation after forming, as shown in Figure 4.15. Therefore, there is a need to develop new testing methods to understand the role of anelastic creep recovery and how it influences total deformation and springback after creep age forming.

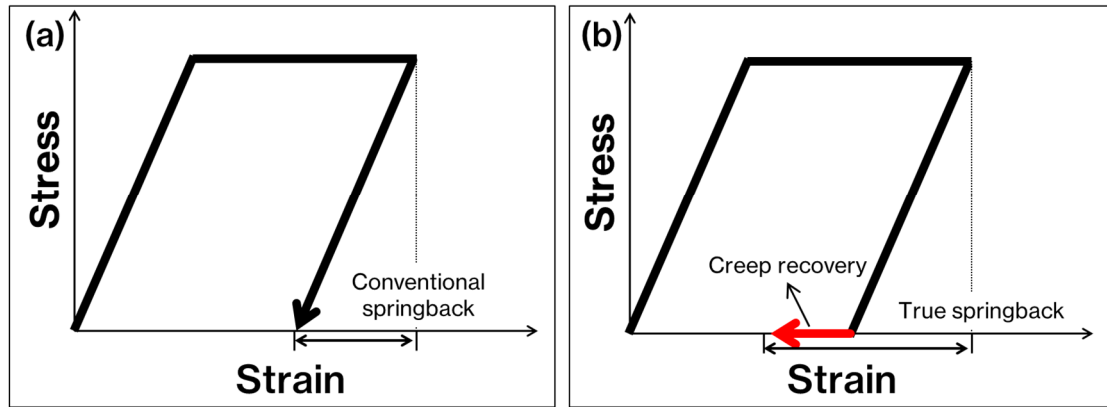


Figure 4.15. (a) Springback determined by conventional creep aging test. (b) True springback determined by CART.

In creep age forming applications, full anelastic creep recovery in strain may never occur in the forming process because the time for full recovery is much longer than the time for loading and unloading. Furthermore, after the forming process, the formed parts are unloaded and cooled to room temperature, at which anelastic creep recovery might not easily take place. Under such circumstances, the ‘drive’ that causes anelastic strain is stored in the formed parts in the form of ‘internal stresses’.

4.3.2 Effect of anelastic creep recovery on total creep

Sinha [1] developed the Strain Relaxation, creep, and Recovery Test (SRRT) for evaluation of the viscous glide and climb controlled dislocation creep in polycrystalline ice. SRRT is capable of determining both the time dependent reversible (anelastic) strain, ε_{an} , and irreversible (viscous) strain, ε_v , of primary creep.

Based on SRRT, with modification to the creep aging condition for age hardened alloys, the creep aging and recovery test (CART) was conducted on AA7050-TAF alloy at 174°C for durations of 2, 4, 6, and 8h under different stress levels of 137.5, 150, and 162.5 MPa, as shown in Figure 4.16. The curves in these graphs show an instantaneous elastic response and typical creep strain-time response when the material is under an applied load. During unloading, all the curves show instantaneous elastic recovery, time dependent anelastic creep recovery, and permanent viscous creep strain after unloading.

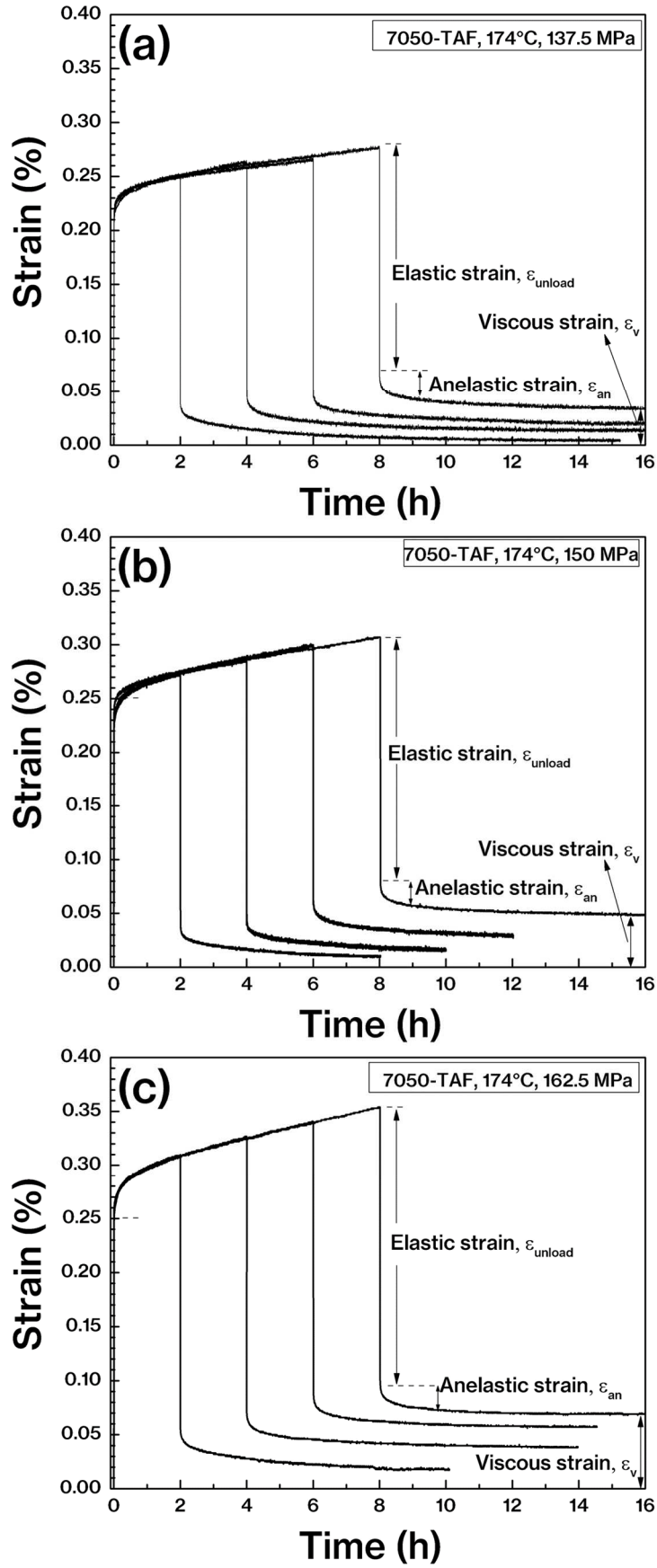


Figure 4.16. CART curves of AA7050-TAF alloy for creep aging times of 2, 4, 6, and 8h under applied stresses of (a) 137.5MPa, (b) 150 MPa, and (c) 162.5 MPa.

Figure 4.17 compares CART curves under three stress levels of 137.5, 150, and 162.5MPa at 174°C for 8h. As expected, increases in applied stress from 137.5 to 150 and 162.5 MPa increase the amount of viscous strain. The anelastic strain in Figure 4.17 at the end of elastic unloading and the beginning of anelastic creep recovery was determined by plotting stress-strain diagrams of the corresponding CARTs. In order to make stress-strain diagrams available, the changes in applied stress were recorded with time throughout the testing. This is a notable advantage of the CART over conventional constant load creep aging tests.

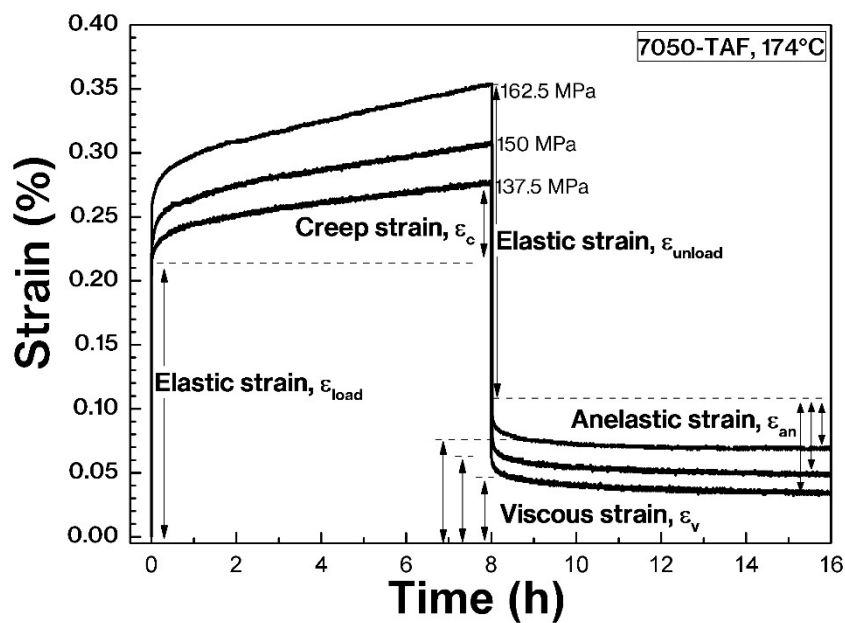


Figure 4.17. Creep aging and recovery curves for creep aging time of 8h under stresses of 137.5, 150, and 162.5MPa at 174°C.

Figure 4.18 shows a typical CART stress-strain diagram for the tests of 8h creep aging time at 174°C under the stress of 162.5MPa, the maximum applied stress in the CART series. A comparison of stress-strain diagrams for the tests of 8h under three stress levels of 137.5, 150, and 162.5MPa at 174°C is shown in Figure 4.19. The stress illustrated in this graph is the average stress captured from the output of the load cell. The linear stress-strain responses in loading and unloading show that loading and unloading are purely elastic, and no plastic strain is induced.

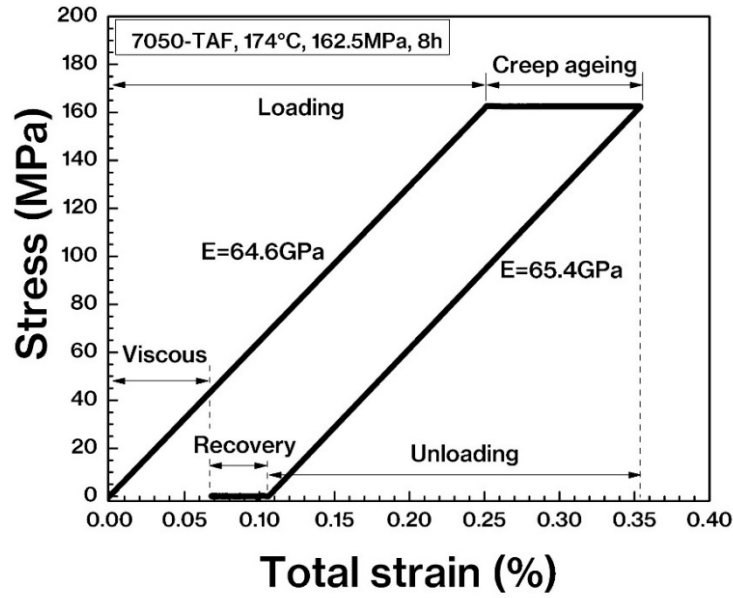


Figure 4.18. Stress-strain diagram showing the hysteresis loop of AA7050-TAF alloy for creep aging time of 8h under stress of 162.5MPa at 174°C.

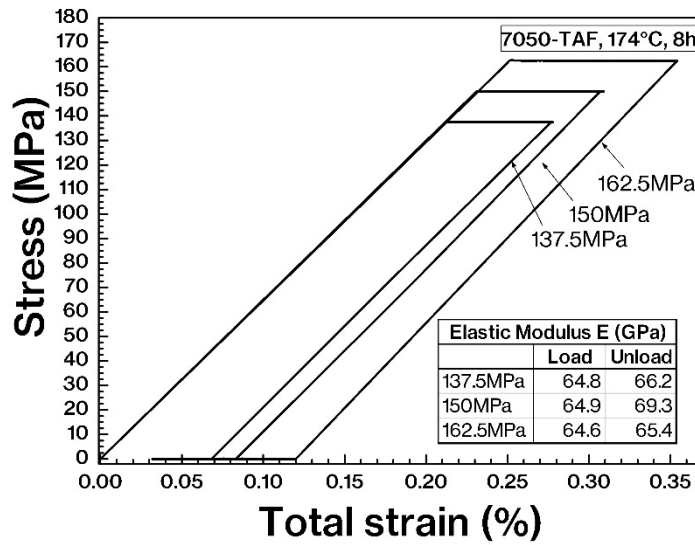


Figure 4.19. Creep aging and recovery hysteresis loop of AA7050-TAF alloy for 8h for applied stresses of 137.5, 150, and 162.5MPa at 174°C.

Two values of Young's modulus, E_{load} and E_{unload} , can be acquired from loading and unloading in the CART. In the case of Figure 4.18, the values of Young's modulus are 64.6 GPa and 65.4 GPa for E_{load} and E_{unload} respectively. There is little difference in the value of Young's modulus in loading and unloading. For a material, the changes in Young's modulus occur only when the stiffness of the precipitates is substantially higher or lower than that of the matrix, and the total volume fraction of precipitates becomes high enough to possess a non-negligible effect [171]. The examination of microstructures under different

pure aging and creep aging conditions, as presented in the previous section, revealed no such difference. This result leads to the conclusion that the effect of creep aging condition on Young's modulus can be neglected.

The recovery part of the hysteresis loop in Figure 4.18 clearly shows that the value of total strain continues to decrease when the stress is equal to zero. The respective amounts of the anelastic strain, ε_{an} , and the viscous strain, ε_v , can be defined from here, whereas conventional constant load creep aging tests can only provide the value of creep strain, ε_c , which is equal to $\varepsilon_{an} + \varepsilon_v$.

Figure 4.20 shows the dependence of the creep strain and the measured ε_{an} and ε_v from each CART in Figure 4.16 on creep aging times of 2, 4, 6, and 8h. The value of ε_{an} appears to remain the same with increases in creep aging time, and ε_v is linear with time.

This trend is in good agreement with the rheological model developed by Sinha [100, 102] for the forged titanium alloy Ti-6246 and nickel-based superalloy Waspaloy; in that model, the recovered strain increases rapidly with time in the primary creep stage, and then the rate of increase decays with time until it reaches a saturated value in secondary creep. According to this model, the total strain can be decomposed into three distinct components:

$$\begin{aligned}\varepsilon_t &= \varepsilon_e + \varepsilon_{an} + \varepsilon_v \\ &= \frac{\sigma}{E} + c \left[\frac{(\sigma - \sigma_0)}{E} \right]^S \ln(t + 1) + t \dot{\varepsilon}_{v(u)} \left(\frac{\sigma}{\sigma_u} \right)^{n(v)}\end{aligned}$$

where the first term ε_e stands for the instantaneously reversible elastic strain; the second term ε_{an} represents the anelastic strain; the third term ε_v is related to the irreversible (viscous) strain; c is the structurally dependent material parameter at constant temperature; and S is a stress exponent. σ_0 is a threshold stress below which no anelastic strain will occur. Young's modulus, E , is used in this term to indicate that anelasticity is related to the stored elastic energy within the material during creep. ε_v is caused by the motion of dislocations, which causes permanent strain, where σ_u is a unit stress (1 MPa) and $\dot{\varepsilon}_{v(u)}$ is the equivalent viscous strain rate for unit stress.

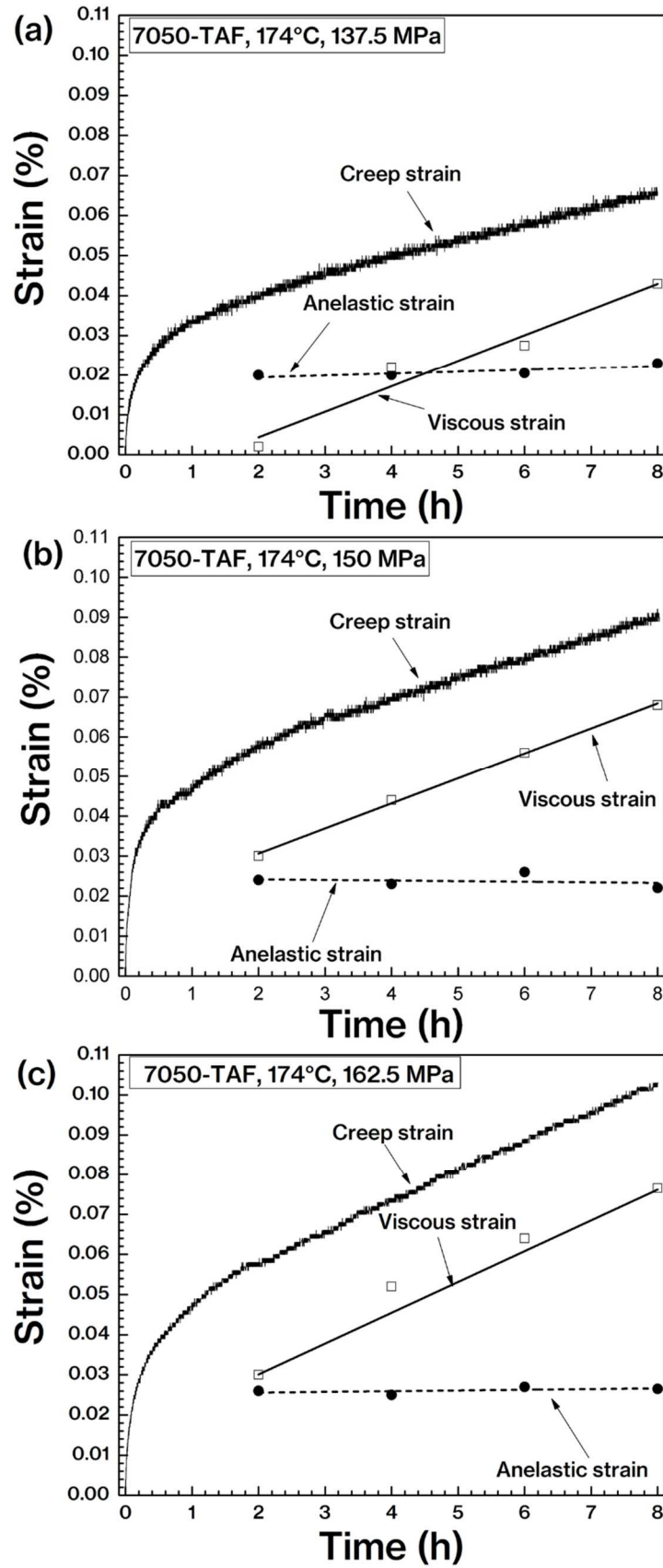


Figure 4.20. Creep strain for the 8h test and the corresponding viscous and anelastic strains for creep aging times of 2, 4, 6 and 8h under applied stresses of (a) 137.5MPa, (b) 150 MPa, and (c) 162.5 MPa.

From the experimental results and the modelling work by Sinha [3], it is suggested that anelastic creep strain after unloading, has a significant effect on total strain. For example, the anelastic strain may contribute up to 30% of the creep strain after 8h of creep aging. This also implies that the total deformation determined through conventional creep aging tests can be over-predicted for the actual creep age forming process.

Since the CART can provide complete data on total deformation from loading to unloading, it is more suitable than the conventional creep tests, which do not consider the change in total strain after unloading, for investigating total amount of permanent deformation after creep age forming. It therefore has the potential to become a new standard experimental method for the development of constitutive equations for creep age forming.

Most studies in the field of creep age forming have generally treated creep strain ε_c as permanent deformation and not considered the anelastic strain ε_{an} in creep age forming. This is the weakness of conventional tests in estimating the final shape of the formed parts in creep age forming. Through the descriptions above, it is clear that the CART can distinguish reversible elastic, anelastic, and viscous (permanent) strain. This distinction provides more benefits than conventional testing methods for creep age forming technologies.

4.3.3 The back stress effect after creep aging

The central cause of anelastic creep recovery after unloading from creep is the existence of internal stresses inside the material. Many experimental methods have been developed to measure the internal stresses in creep for different materials, and an overview of these tests has been presented in Chapter 2. This section presents a new testing technique developed to characterise the role of internal stresses on anelastic creep recovery. The test is called the creep aging and back stress test (CABT). The term *back stress* is used to represent the internal resistance that arises in the material during forming. It is defined as an imaginary force arising from the prior deformation history or from the deformation process itself [95].

The basic concept behind the development of this testing method is to determine the stress response of the material when the strain is held constant after unloading from creep, where

anelastic creep strain recovery is impeded. In the creep aging process, the back stress acts in the opposite direction of the external applied force. The materials, therefore, are only subjected to an effective stress – which is the difference between applied stress and internal stress under the creep aging condition. When the external force is removed, the internal stress continues to work on the material in the opposite direction of the external force and causes anelastic creep recovery. If the strain of the material is held constant after the removal of the external force to prevent creep recovery, the back stress that causes anelasticity can be expected to remain inside the material.

The method of performing a CABT is to apply a constant stress on the material at an elevated temperature for the conventional creep test in the first step. After creep for a certain time, the stress is removed, and the strain of the material is held constant to monitor the evolution of stress generated from the material.

The efficacy of CABT is described in Figure 4.21, which shows the stress-time diagram of a CABT with the combination of the strain-time diagram of CART for the test of 6h of creep aging under an applied stress of 150 MPa. The scale of stress and strain in this graph follows the stress strain relationship, $\sigma = E \cdot \varepsilon$, where $E=65.2$ GPa as the stress on the scale bar is below 150 MPa. Therefore, the curves of the CART and CABT are comparable. As shown in the graph, after unloading from the constant stress in creep at 6h, an increase in stress with time can be observed when the strain is held constant.

The result of the CABT is comparable to the strain-time diagram of the CART under the same applied stress of 150 MPa and creep aging time of 6h. As the response of strain in the CART generally decreases with time, the response of stress in the CABT gradually increases with time. The values of strain and stress in the CART and CABT show a trend to merge.

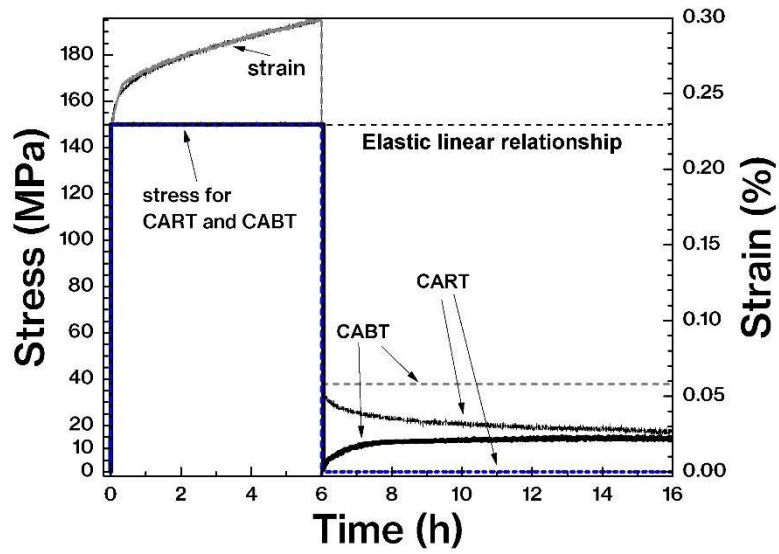


Figure 4.21. Relationship between the creep strain recovery from CART and the back stress induced from CABT for the AA7050-TAF alloy under the applied stress of 150 MPa for 6h.

Figure 4.22 presents a stress-time diagram of the CABT for 5, 10, and 30 minutes and for 1, 2, 4, 6, and 8h at stress of 162.5 MPa. A trend can be observed; the back stress increases with the creep aging time, and the stress stops increasing when the creep aging time reaches a certain level.

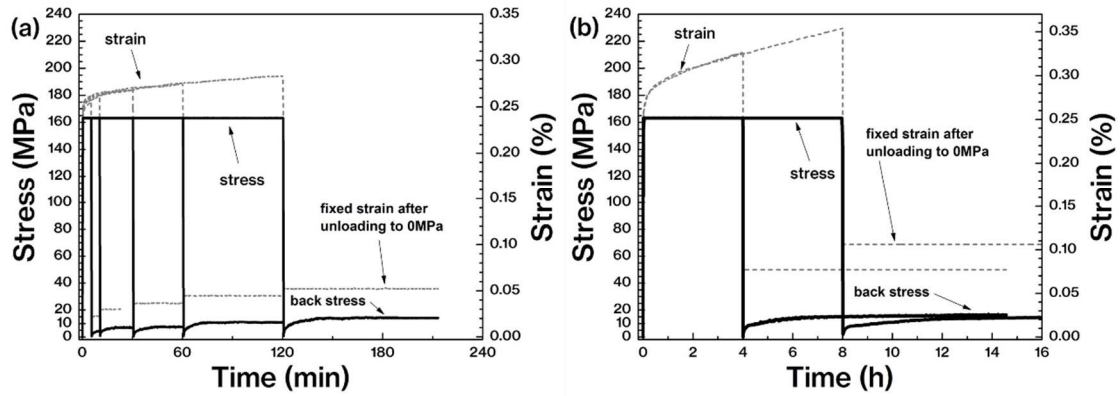


Figure 4.22. Creep aging and back stress curves under stress of 162.5 MPa at 174°C, where the strain is held as a constant after 5 min, 10 min, 30 min, 1h, 2h, 4h, 6h, and 8h of creep aging, respectively.

The maximum values of the back stress obtained from all the CABTs under the stress levels of 137.5, 150, and 162.5 MPa are plotted versus creep aging time in Figure 4.23. The results indicate that the maximum back stress increases rapidly with time up to 2h and then gradually decelerates to a saturated value afterwards. As stated in section 2.3.1, Sinha

observed that the recoverable anelastic strain increase rapidly with time in the primary creep stage and reach a saturation level in the secondary creep stage. Because the anelastic strain and back stress follows stress strain relationship, $\sigma = E \cdot \varepsilon$, as shown in Figure 4.21, the anelastic strain and back stress are interchangeable by multiplying Young's modulus, E . The result in Figure 4.23 is, therefore, consistent with the observation by Sinha [100, 102].

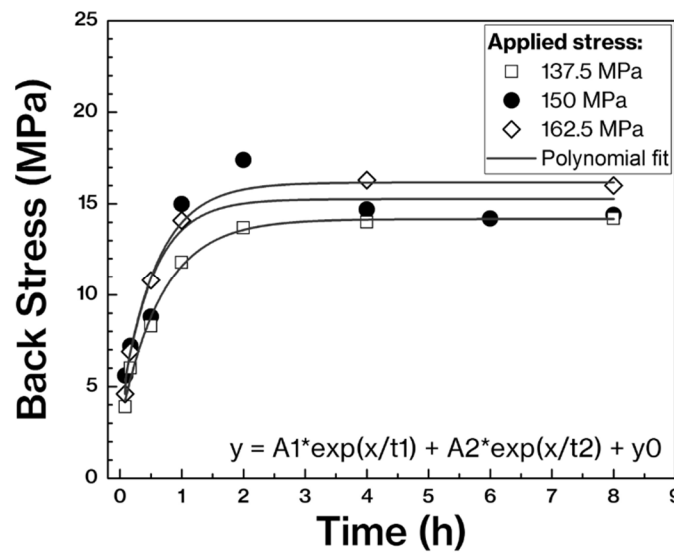


Figure 4.23. Time dependence of back stress for AA7050 alloy at 174°C derived from CABTs under applied stress levels of 137.5, 150, and 162.5 MPa.

4.3.4 Mechanisms behind anelasticity

The experimental work from CABTs revealed that anelasticity is related to primary creep. Primary creep can be described in terms of two strain components [102]: irreversible viscous (permanent) strain and reversible anelastic strain. In the early stage of primary creep, the viscous creep rate is constant and the anelastic strain follows a logarithmic decay function [102]. The viscous strain component may be caused by the glide of dislocations, which is thermally activated and stress dependent, whereas reversible anelastic strain has been ascribed to the slide and shear of grain boundaries, which store energy inside the material [102], or the existence of back stress, σ_b , which works in the opposite direction of the deformation, [95, 172, 173]. The creep process is therefore controlled by the effective stress ($\sigma - \sigma_b$). σ_b is characterised by the substructure of the material, such as subgrains or dislocation cell structures that develop during the primary creep stage [127].

TEM was performed to observe the dislocation cell structures. Figure 4.24 shows the dislocation cell structure of the AA7050-TAF alloy after a 12h CABT under 137.5 MPa and an 8h CART under 150 MPa. Well-developed dislocation cell structures can be found in these samples. The two-dimensional dislocation cell structures has been proposed [106] as the schematic diagram shown in Figure 2.6. Since during creep deformation the dislocations are difficult to break through the cell walls, they tend to pile up along the walls and result in a higher shear stress on the cell walls (hard regions), whereas very little or no dislocations can be found within the cell interiors (soft regions). When the material is unloaded from creep in tension, the stress in the hard region is still positive but the stress in the soft region becomes negative, as shown in Figure 2.7.

However, the dislocation microstructure developed during creep aging is very unstable that any changes in the stress conditions may affect its morphologies. The cell structure observed here may only reflects the remainder of the immobile dislocations, where those mobile dislocations may dissipate or rearrange into a lower-energy configuration due to unloading to stress-free condition. Only two samples, therefore, were selected for TEM investigation. Further investigation of dislocation cell structures, including size and distribution, in the related samples is required in order to explore the effect of dislocation arrangement on the anelasticity.

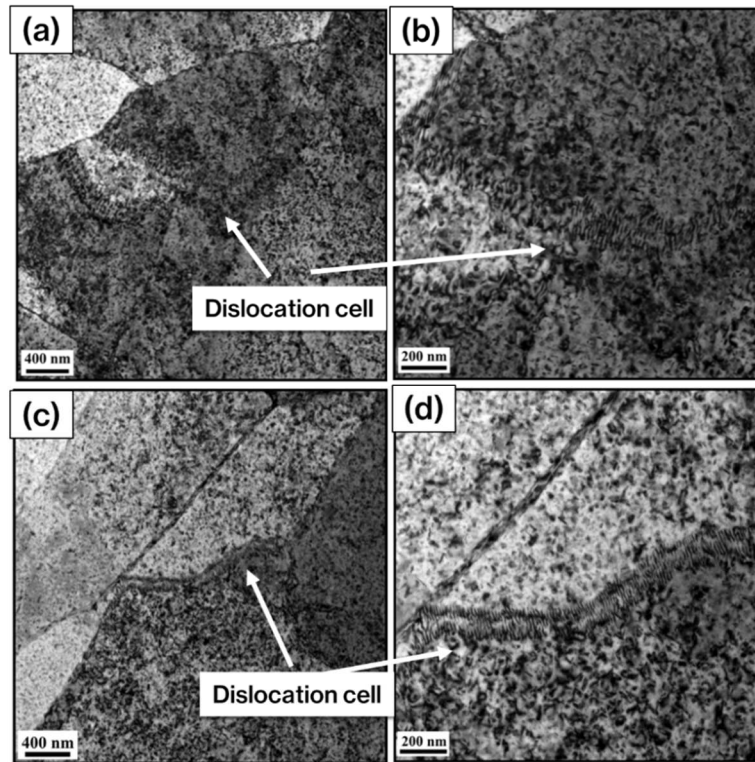


Figure 4.24. Dislocation cell structures of the AA7050-TAF alloy (a-b) after 12h CABT for 137.5 MPa and (c-d) after 8h CART for 150 MPa.

4.3.5 Summary

Two experimental methods, the CART and CABT, have been performed. The important results of the CART are summarised as follows.

- CART provides more complete data than conventional creep aging tests for predicting springback, since springback also includes anelastic creep recovery.
- The increase in creep aging time from 2 to 8h does not affect the anelastic strain generated by anelastic creep recovery, whereas the viscous strain increases with time. This result is consistent with the research by Sinha [102], wherein primary creep can be described in terms of two strain components: irreversible viscous (permanent) strain and reversible anelastic strain. This leads to the conclusion that the anelastic creep recovery effect is independent of secondary creep.
- The recoverable strain increases from about 0.02% to 0.026% when the applied stress increases from 137.5 to 162.5 MPa. It is concluded that the anelastic creep recovery is proportional to the applied stress.

CABTs have been performed to identify the back stress of anelastic creep recovery. The important CABT results can be summarised as follows.

- The back stress increases with the applied stress, and the result is comparable to the CART results. There is a direct relationship between anelastic creep recovery and back stress.
- The back stress increases in the primary creep stage, and then it remains constant in the secondary creep stage.

As explained in the literature review in Chapter 2, back stress can originate in the subgrain cell structures formed by mobile dislocations. When the material is unloaded from creep, the subgrain cell structures break down, and the distribution of mobile dislocations changes. The dislocation cell structures have been observed by TEM microstructural analysis.

4.4 The effect of anelasticity on springback in the three-point bending test

In the uniaxial creep aging and relaxation tests (CART) and creep aging and back stress tests (CABT), the anelasticity of the alloy has been demonstrated to influence the total strain after creep aging. In the actual forming process, however, the distribution of the applied stress on the forming part is more complicated than that in the uniaxial loading condition. The forming part is subjected to bending, which produces compression in one half and tension in the other half. The anelastic behaviour of the forming parts under this condition is still unclear. The three-point bending stress relaxation and back stress test (SRBT) is therefore designed to examine the relationship between the thickness of the material and the amount of anelastic creep recovery after creep age forming. This section aims to interpret the results from this test.

4.4.1 Deflection and springback

All the symbols used in the SRBT series are given in Table 4.3, and the overall experimental data are listed in Table 4.4. The experimental curves of load relaxation and recovery with their corresponding deflections of the SRBT for three thicknesses are shown in Figure 4.25. This graph indicates that under the three-point bending condition, the load relaxation behaviour is similar to the stress relaxation stress-time curves under the uniaxial condition. Based on these experimental data, four parameters – load relaxation ratio

(R_{load}), instant springback ($SP_{instant}$), percentage of deflection recovered ($R_{deflection}$), and true springback ($SP_{recovery}$) – can be defined. The load relaxation ratio (R_{load}) is given by

$$R_{load} = F_{end} / F$$

where F is the initial applied load of the SRBT and F_{end} is the residual load after 8h of stress relaxation before first stage unloading. This parameter provides an indication in how much load can be relaxed during the forming process.

Table 4.3. Symbols used in SRBTs.

L (mm)	Length of the material	d_p (mm)	Thickness of the material
F (N)	Applied load	F_r (N)	Load recovered during back stress stage
F_{end} (N)	Residual load before unloading	R_{load} (%)	Load relaxation ratio F_{end}/F
D (mm)	Maximum deflection	$SP_{instant}$ (%)	Instant springback $(D - D_{unload})/D$
D_{unload} (mm)	Maximum deflection after unloading	$R_{deflection}$ (%)	Percentage of deflection recovered $(D_{unload} - D_r)/D_{unload}$
D_r (mm)	Maximum deflection after unloading from load recovery	$SP_{recovery}$ (%)	True springback considering recovery $(D - D_r)/D$

Table 4.4. Experimental data of SRBTs.

Unit test	thickness (mm)	F (N)	F_{end} (N)	σ_{max} (MPa)	D (mm)	D_{unload} (mm)	F_r (N)	D_r (mm)
1	3	100.3	70.5	97.6	2	0.84	12	0.75
2	3	366.7	231.4	356.6	5.47	2.09	26.7	1.84
3	3	445	306.1	432.7	7.88	3.75	36.1	3.25
4	5	366.8	245.7	128.4	1.28	0.38	23	0.31
5	5	512.4	383.4	179.4	2	0.64	37.5	0.53
6	5	744.8	520.2	260.7	3	0.89	48.4	0.69
7	5	1058.3	739.6	370.5	4	1.3	65.1	1.01
8	5	1300	836	455	5	1.84	81.9	1.47
9	8	1058.3	846	144.7	1.1	0.25	60.4	0.19
10	8	2193.2	1631.4	299.9	2	0.52	102.4	0.4
11	8	3256.4	2202.1	445.3	3.3	1.21	188.5	1.01

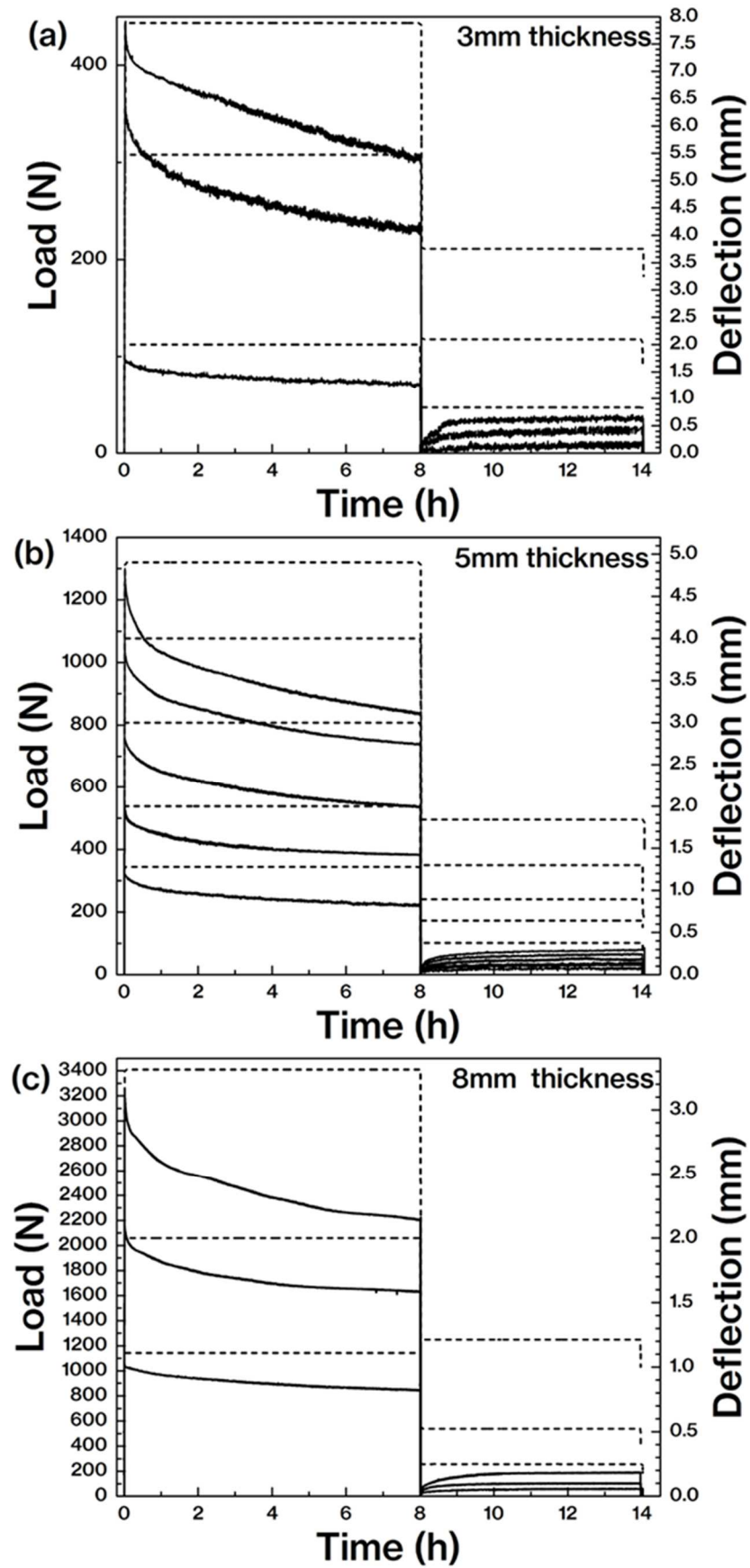


Figure 4.25. Load relaxation (solid lines) and deflection (dashed lines) curves of (a) 3mm, (b) 5mm, and (c) 8mm thick material.

The instant springback index [159], $SP_{instant}$, is defined as below:

$$SP_{instant} = (D - D_{unload})/D$$

where D is the maximum deflection of the formed part after loading, and D_{unload} is the maximum deflection after first stage unloading. This equation is used to compare the shapes of the formed part after loading to a certain deflection and after instant springback. The percentage of deflection recovered after load recovery is defined as follows:

$$R_{deflection} = (D_{unload} - D_r)/D_{unload}$$

where D_r is the maximum deflection after second stage unloading. As mentioned previously, the SRBT involves a secondary load recovery stage, where the deflection of the formed part is held constant after unloading to zero load. At this stage, due to the inhibition of strain recovery, anelastic creep recovery appears in the form of load recovery, similar to the back stress recovery presented in Section 4.3.3. For the formed parts of three thicknesses, the higher the applied load is, the higher the back stress is. Full unloading was performed in each SRBT to fully remove the load recovered in the second stage. By considering load recovery, the total springback index $SP_{recovery}$ can be defined as follows:

$$SP_{recovery} = (D - D_r)/D$$

Table 4.5 shows the calculated results based on the experimental data in Table 4.4 and the equations above. The calculated results show that the values of R_{load} in all SRBTs of three thicknesses are between 25-36%. For formed parts of 3mm and 5mm thickness, there is no obvious trend of how the thickness and the deflections could affect R_{load} . For formed parts of 8mm thickness, the higher the deflection, the higher the R_{load} .

Table 4.5. Calculated data of SRBTs based on Table 4.4.

	thickness (mm)	<i>D</i> (mm)	<i>R_{load}</i> (%)	<i>SP_{instant}</i> (%)	<i>R_{deflection}</i> (%)	<i>SP_{recovery}</i> (%)
1	3	2	29.8	58	11	63
2	3	5.47	36.9	62	13	67
3	3	7.88	31.3	53	14	59
4	5	1.28	33.1	71	19	76
5	5	2	25.2	69	18	72
6	5	3	30.2	71	23	77
7	5	4	30.2	68	23	75
8	5	4.9	35.7	63	21	71
9	8	1.1	20.1	78	24	83
10	8	2	25.7	74	24	80
11	8	3.3	32.4	64	17	70

R_{load} can also mean the degree of deformation; a higher R_{load} can therefore mean lower instant springback, $SP_{instant}$, of the formed parts. The $SP_{instant}$ of the formed parts of 3mm thickness is between 53-62%, which is generally lower than the $SP_{instant}$ of formed parts of other thicknesses, where the value is between 63-78%. $SP_{instant}$ of formed parts of 5mm and 8mm thickness generally decrease with the increasing deflections.

For the percentage of deflection recovered after load recovery, $R_{deflection}$, the values of formed parts of 3mm thickness are about 11-14%, which are generally lower than the $R_{deflection}$ of formed parts of other thicknesses, the values of which are about 17-24%. When considering the effect of load recovery on total deformation, true springback, $SP_{recovery}$, is 5-7% higher than $SP_{instant}$.

4.4.1.1 Forming under the same maximum flexural stress

The maximum deflections for the SRBTs are determined by maximum flexural stress. The value of maximum flexural stress was set at 460 MPa, which is just below the high temperature yielding point of the AA7050-TAF alloy used in this study. The SRBT curves of the plates in three thicknesses under this stress level are shown in Figure 4.26. The maximum flexural stress is calculated by classical beam theory, $\sigma_{max} = 3PL/(2bd_p^2)$, where P is the maximum load (N), L is the length (m), b is the specimen width (m), and d_p is the specimen thickness (m).

Comparing the formed parts of three thicknesses under the same maximum flexural stress (460 MPa), as presented in Table 4.6, the deflection of the formed parts of 3mm thickness

reaches 7.88mm, whereas those of the formed parts of 5mm and 8mm thickness are 4.9mm and 3.3mm. The springback of formed parts of 3mm thickness, however, is much lower than that of formed parts of 5mm and 8mm thickness.

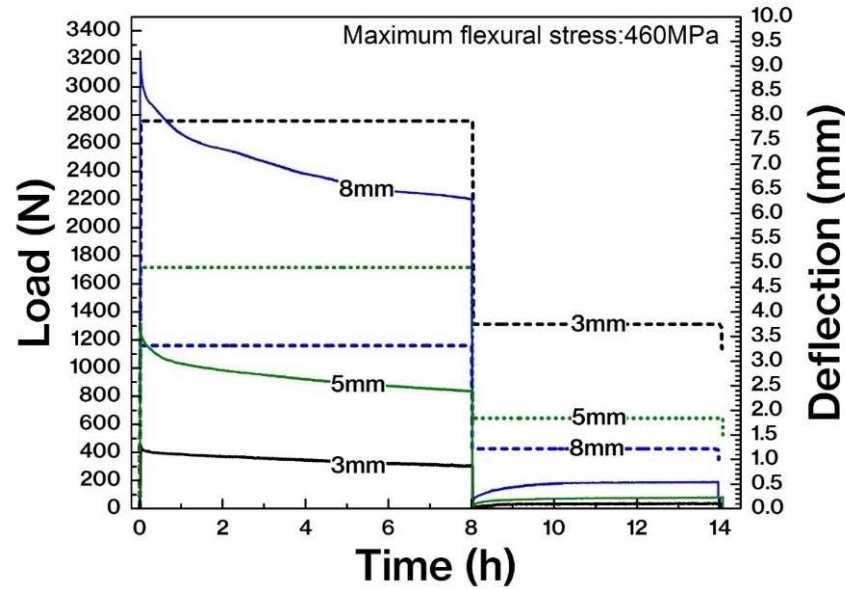


Figure 4.26. Load relaxation and displacement curves of the materials of 3mm, 5mm, and 8mm thickness under the same flexural stress.

Table 4.6. Data of SRBT on samples under maximum flexural stress of 460MPa.

Sample No.	thickness (mm)	R_{load} (%)	D (mm)	$SP_{instant}$ (%)	$R_{deflection}$ (%)	$SP_{recovery}$ (%)
3	3	31.3	7.88	53	14	59
8	5	35.7	4.9	63	21	71
11	8	32.4	3.3	64	17	70

The cause of this anomaly is that the deflection is too large in the 3mm case. The contact surface area of the sample and the three-point bending rig might increase due to large deformation. The flexural stress calculated from the classical beam model is therefore inaccurate. In classical beam theory, the assumptions are that beam deflections are small, and the cross-section remains normal to the deformed axis of the beam [174].

4.4.1.2 Forming under the same deflection

Figure 4.27 presents the SRBT curves of the formed parts of three thicknesses loaded to the same deflection (2mm). For the formed part of 8mm thickness, it took a significantly higher load, 2193.2 N, to deform the part to 2mm deflection, whereas it only took 100.0 N and 512.4 N to deform the parts of 3mm and 5mm thickness respectively. The higher

applied load results in higher $R_{deflection}$, as shown in Table 4.7. A similar trend can be observed in $SP_{instant}$ and $SP_{recovery}$. These values increase with the thickness. Full details of the comparison are provided in Table 4.7.

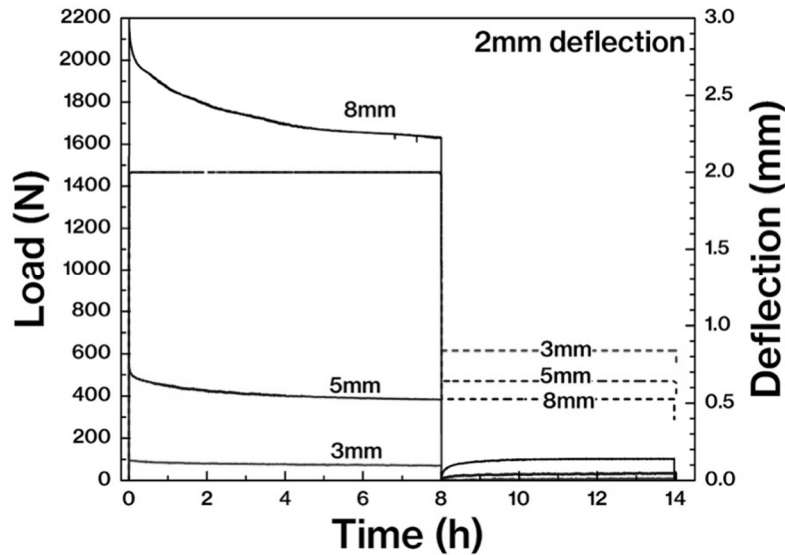


Figure 4.27. Load relaxation and displacement curves of the materials under the same deflection.

Table 4.7. Calculated data of SRBTs on samples under the same deflection of 2mm.

Sample No.	thickness (mm)	R_{load} (%)	D (mm)	$SP_{instant}$ (%)	$R_{deflection}$ (%)	$SP_{recovery}$ (%)
1	3	29.8	2	58	11	63
5	5	25.2	2	69	18	72
10	8	25.7	2	74	24	80

4.4.1.3 Formed parts under same applied load

Figure 4.28 shows the load relaxation and recovery curves of the formed parts under the same applied load. Under the applied load of 366.7 N, the load of the formed parts of 3mm thickness relaxed faster than did that of parts of 5mm thickness, as shown in Figure 4.28a. A similar trend can be observed in Figure 4.28b, where under the applied load of 1058.3 N, the load relaxation ratio of 5mm formed parts is higher than that of 8mm formed parts. Full details of the comparison are provided in Table 4.8 and Table 4.9.

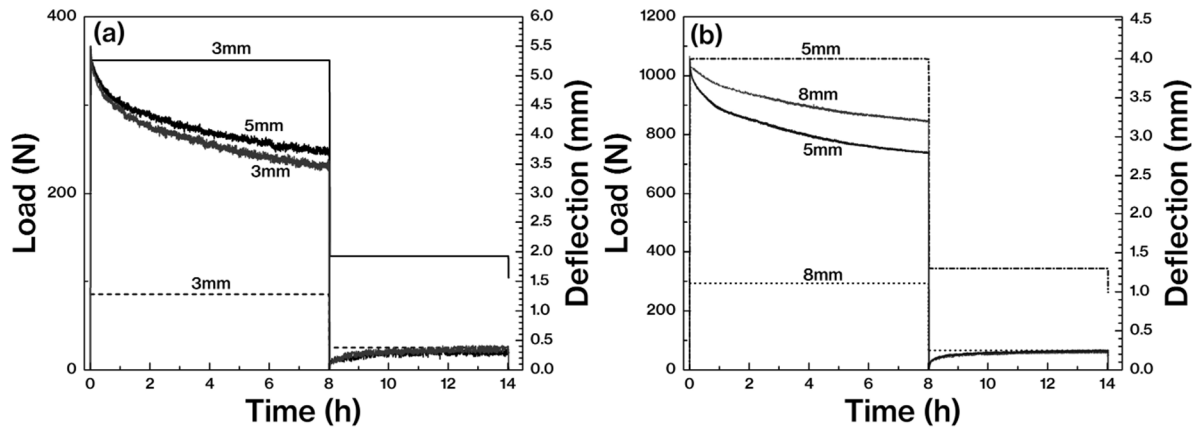


Figure 4.28. Load relaxation and recovery curves of the formed parts under the same applied load, (a) 3mm versus 5mm; (b) 5mm versus 8mm.

Table 4.8. Calculated data of SRBTs on samples under the same applied load (F) of 366.7 N for 3mm and 5mm thick plate.

Sample No.	thickness (mm)	F (N)	R_{load} (%)	D (mm)	$SP_{instant}$ (%)	$R_{deflection}$ (%)	$SP_{recovery}$ (%)	F_r (N)
2	3	366.7	36.9	5.47	0.62	13	67	26.7
4	5	366.8	33.1	1.28	0.71	19	76	23

Table 4.9. Calculated data of SRBTs on samples under the same applied load (F) of 1058.3 N for 5mm and 8mm thick plate.

Sample No.	thickness (mm)	F (N)	R_{load} (%)	D (mm)	$SP_{instant}$ (%)	$R_{deflection}$ (%)	$SP_{recovery}$ (%)	F_r (N)
7	5	1058.3	30.2	4	0.68	23	75	65.1
9	8	1058.3	20.1	1.1	0.78	24	83	60.4

4.4.2 Summary

Three-point bending stress relaxation and back stress tests (SRBTs) have been conducted on the AA7050-TAF alloy for 8h for stress relaxation and 6h for load recovery at 174°C. Since the load applied on the formed part was constantly monitored by the load cell of the testing machines, the load relaxation and recovery behaviours were monitored and characterised. The findings can be summarised as follows:

- When the same maximum flexural stress of 460 MPa is applied on material of three thicknesses, the amount of springback is about the same (70%) for specimens of 5mm and 8mm thickness. For specimens of 3mm thickness, springback is much lower (59%).
- The thinner the sheets of metal are, the higher the stress relaxation rate. This leads to lower springback on thinner material.

- The percentage of deflection recovery caused by back stress, $R_{deflection}$, is about 11-14% for formed parts of 3mm thickness, and 20-24% for formed parts of 5mm and 8mm thickness. The load recovery contributes about 5-7% of the springback in general.

4.5 Concluding remarks

An experimental investigation has been conducted on the AA7050-TAF alloy. The creep aging behaviour was examined with the fundamental mechanical testing presented in section 4.1. The results from mechanical testing generally show typical overaging behaviour, which is dependent on not only time but also the applied stress. Microstructure characterisation was then performed with Transmission Electron Microscopy (TEM) to investigate the micro-mechanisms behind creep aging. It has been shown that η' precipitates are the main contributor to the strength of the material. Although applying stress can promote the growth rate of GP zones and η' precipitates, it also speeds up the overaging process, wherein the coarsening of η precipitates softens the material.

Since the main interest in creep age forming is the final shape of the material, two experiments – the creep aging and recovery test (CART) and the creep aging and back stress test (CABT) – have been developed to investigate how anelastic creep recovery can affect the final deformation of the material. The results show that the back stress gradually develops during the primary creep stage. TEM microscopy has been performed to investigate the effects of dislocation cell structures on anelasticity. Although these structures have been identified, efforts still need to be made to understand how they contribute to anelasticity.

Three-point bending stress relaxation and back stress tests (SRBTs) have been performed to simulate the creep age forming process. With the implementation of the load cell on the test machine, load relaxation behaviour can be monitored during the forming process; this approach has yet to be applied in other studies or in industrial practice. The experimental data in this chapter will be used for building material models for FE implementation.

5 UNIFIED CONSTITUTIVE MODELLING

This chapter details the development of constitutive modelling techniques from a multitude of theories and experiments in creep aging and anelastic creep recovery. It will delineate the basic principles that facilitate microstructural variations in the creep age forming process. In doing so, it will present an overview of some of the micro-mechanisms already covered in Chapter 2, so as to develop an integrated statement for the construction of a new constitutive model.

5.1 Introduction

It is of greatly increasing interest to develop constitutive models that are solidly based on deformation mechanisms. Mechanism-based constitutive equations are, indeed, very attractive in the fields of metal forming technologies because they create the ability not only to describe and predict the mechanical behaviour in and after the forming process, but also to characterise both deformation and strengthening mechanisms, which involve the forming process.

The physical mechanisms are applied in mechanism-based constitutive models quite differently from how they are applied in ‘first principle’ models. ‘First principle’ models apply the fundamental physical mechanisms in quantitative equations in which each variable stands for a different actual physical entity. However, this approach, the root of

the science, has some limitations when it comes to material modelling – the mechanisms or the physical entities of the materials are too complicated to be represented in the equations, and the ‘first principle’ theories that can predict the mechanical behaviour or micro-mechanisms of the material are not yet available.

Mechanism-based constitutive models, on the other hand, focus on the general trends of the mechanical behaviour and mechanisms of materials, which can be observed from experiments or proposed in theories. The equations possess a combination of physical mechanisms and empirical phenomenological data acquired from experiments.

The basic concept of a mechanism-based constitutive model is to apply internal state variables to characterise the structural state of the material and to formulate the constitutive equations that govern the internal state variables so that the mathematical trend of the internal state variables can replicate the mechanical behaviour and mechanisms in the actual materials.

In this chapter, by virtue of internal state variables, the constitutive equations developed by Zhan et al. [48] will be applied and expanded so as to model the over-aging behaviour of the AA7050-TAF alloy, and then a new set of equations will be presented for modelling the inelastic/anelastic behaviour of the material.

5.2 Constitutive modelling for the peak-aged condition

The set of constitutive equations applied in this section was proposed by Zhan et al. [48], who first considered the yield strength evolution of aluminium alloy in terms of three state variables: solid solution strengthening, dislocation hardening, and precipitation hardening. The use of these state variables enables the constitutive equations to model the creep aging behaviour of aluminium alloy to reach an overaged condition. Extensive experiments have been motivated by Zhan’s research [49, 50, 160, 175]. In general, two experiments, uniaxial creep tests and tensile tests, are required to determine the material constants of the constitutive equations based on the curve fitting technique.

5.2.1 Zhan’s unified uniaxial creep-aging constitutive equations

The set of constitutive equations proposed by Zhan is presented as follows:

$$\dot{\epsilon}_c = A_1 \sinh\{B_1[|\sigma|(1 - \bar{\rho}) - k_0\sigma_y]\} \text{sign}\{\sigma\} \quad (5.1)$$

$$\sigma_y = \sigma_{ss} + \sqrt{\sigma_A^2 + \sigma_{dis}^2} \quad (5.2)$$

$$\dot{\sigma}_A = C_A \dot{\bar{r}}^{m_1} (1 - \bar{r}) \quad (5.3)$$

$$\dot{\sigma}_{ss} = C_{ss} \dot{\bar{r}}^{m_2} (\bar{r} - 1) \quad (5.4)$$

$$\dot{\sigma}_{dis} = A_2 \cdot n \cdot \bar{\rho}^{n-1} \dot{\bar{\rho}} \quad (5.5)$$

$$\dot{\bar{r}} = C_r (Q - \bar{r})^{m_3} \cdot (1 + \gamma_0 \bar{\rho}^{m_4}) \quad (5.6)$$

$$\dot{\bar{\rho}} = A_3 (1 - \bar{\rho}) |\dot{\epsilon}_c| - C_p \bar{\rho}^{m_5} \quad (5.7)$$

Equation (5.1) describes the creep strain evolution. Creep strain rate, $\dot{\epsilon}_c$, is a function of applied stress during creep, σ , normalised dislocation density, $\bar{\rho}$, and the material's yield strength, σ_y . σ_y is determined from the state of age hardening, σ_A , solid solution hardening, σ_{ss} , and dislocation hardening, σ_{dis} , which are presented in equations (5.2) to (5.5). These hardening effects are influenced by the normalised precipitate size, $\bar{r}$, and normalised dislocation density, $\bar{\rho}$. The growth rate of the normalised precipitate size, $\dot{\bar{r}}$, is related to the normalised dislocation density, $\bar{\rho}$. The rate equations of these two variables are presented in (5.6) and (5.7). The normalised precipitate size, $\bar{r}$, is defined as

$$\bar{r} = \frac{r}{r_c} \quad (5.8)$$

where r is the original precipitate radius and r_c is the precipitate size in the peak-aged condition. Normalised dislocation density, $\bar{\rho}$, is defined as

$$\bar{\rho} = \frac{(\rho - \rho_{ini})}{\rho_{max}} \quad (5.9)$$

where ρ_{ini} is the dislocation density in the initial state before aging and ρ_{max} is the saturated dislocation density.

This set of equations was originally formulated to model the age hardening behaviour of aluminium alloys that were solution heat treated, or at the T3/T4 condition, but not age softening behaviour. In order to model the softening behaviour due to the coarsening of precipitates in terms of overaging from the peak-aged condition, or the T6/T7 condition, the initial values are determined based on the understanding of the current material, in the

peak-aged condition, in which the precipitates have developed to a certain size and have already contributed the highest strength to the material. In this stage, the strength contributed by solid solution hardening is relatively low because, from an atomic perspective, the alloying elements in the matrix have partly transformed into the precipitates. In view of the above, the initial values for Zhan's equations have been determined as given in Table 5.1.

Table 5.1. Initial values of the constitutive model.

σ_a (MPa)	σ_{ss} (MPa)	σ_{dis} (MPa)	$\bar{\rho}$	$\bar{r}$
300.0	273.4	0.0	0.0	1.0

5.2.2 Calibration and validation of the model from experimental data

The material constants of the constitutive equations were determined by manually fitting experimental creep curve tests at different stress levels and the yield strength value acquired from the tensile test. The fitting process can be divided into three stages:

Stage 1: Determine the material constants related to the normalised precipitate size and dislocation density using experimental results on yield strength evolution with time under the pure aged (0 MPa) and creep aged conditions (137.5, 150, 162.5 MPa), as shown in Figure 5.1.

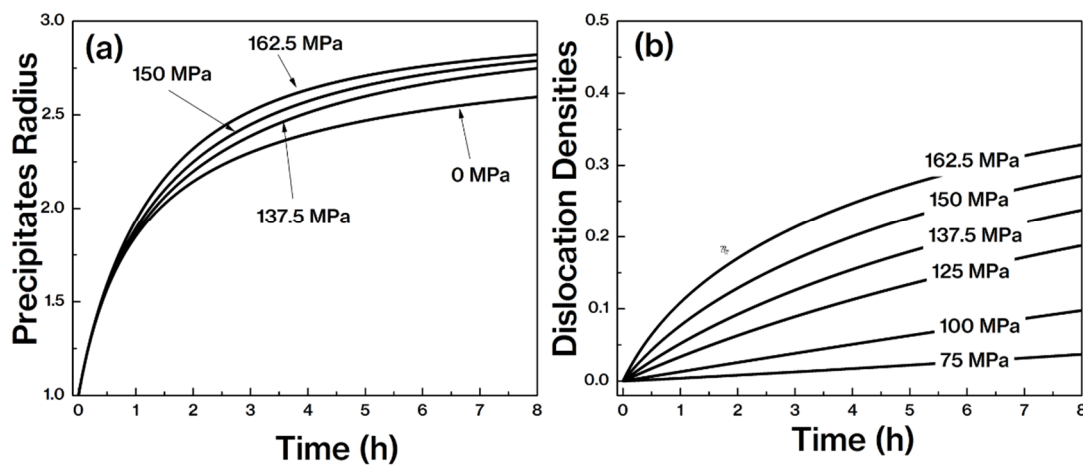


Figure 5.1 (a) Computed normalised precipitate size variation (b) Computed dislocation density variation.

Stage 2: Determination of the material constants related to yield strength using experimental results on yield strength evolution with time under different stress levels, as shown in Figure 5.2a.

Stage 3: Re-evaluation of all pre-determined material parameters according to the experimental creep data and yield strength data, and optimisation of them in Stage 1-2 to obtain the best-fit results, as shown in Figure 5.2b.

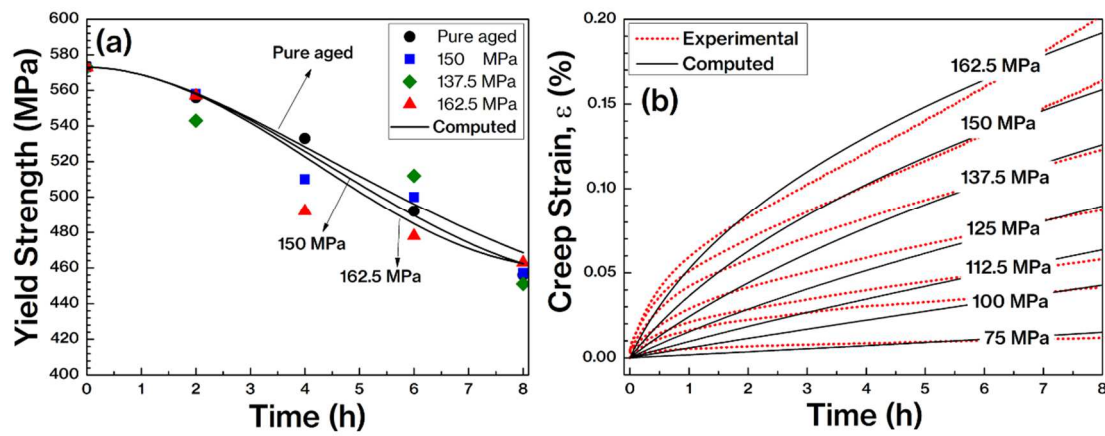


Figure 5.2. (a) Experimental (symbols) and computed (solid lines) yield strength evolution under different stress levels at 174 °C. (b) Experimental (dashed lines) and computed (solid lines) creep-aging curves for different stress levels at 174 °C.

The material constants for constitutive equations determined from the method above are listed in Table 5.2. In order to obtain the material constants efficiently, the fitting process of the evolutionary algorithms (EA) fitting software developed by Zhang et al. [176] was then used to calibrate the pre-determined constants.

Table 5.2. Constants of the constitutive equations for the AA7050-TAF alloy at 174°C.

A_1	B_1	k_0	C_A	m_1	C_{ss}
2.082E-5	0.04	0.097	25	0.5	-1.6
m_2	n	A_2	C_r	Q	γ_0
0.806	1.425	90	0.3	3	50
m_3	m_4	A_3	C_p	m_5	
2.542	2.034	220	0.206	3.312	

Based on the determined material constants, the strength contributions by dislocation hardening, solid solution hardening, and age hardening under different stress levels, 137.5, 150, and 162.5 MPa, can be computed, as given in Figure 5.3a, b, c, and d.

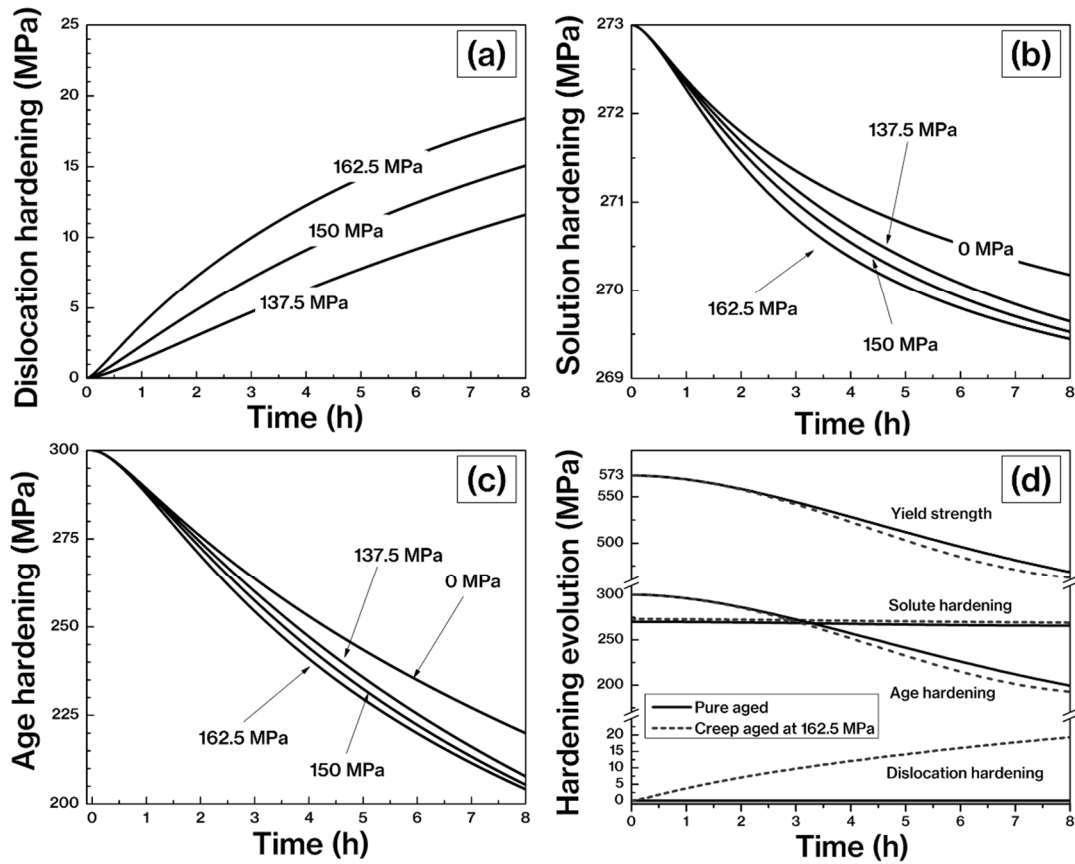


Figure 5.3 Computed state variable of evolution in strength by (a) dislocation hardening, (b) solution hardening, (c) age hardening, and (d) their contribution to yield strength.

5.2.3 Summary

The material constants determined in this section extend Zhan's constitutive equations to model the over-aging behaviour of the aluminium alloy AA7050-TAF. Good agreement can be observed between the experimental and simulated results. The calibration of the material constants of Zhan's constitutive equations can lead to the following results:

- In this constitutive model, the yield strength variation is considered to be contributed by three strengthening mechanisms – solid solution strengthening, dislocation hardening, and precipitation hardening. Both the yield strength and these mechanisms have been successfully modelled.
- The difference in the mechanical properties of the AA7050-TAF alloy under different applied stress conditions is minimised after 8h. This feature is characterised by the model of this work.
- For this peak aged alloy, the coarsening of the precipitates reduces the contribution of precipitation hardening in the over-aging region; this reduction is the main factor in the decrease in the material's yield strength.

5.3 A new model for predicting anelastic creep recovery

Anelastic creep recovery has been demonstrated in the previous chapters to have a significant influence on total springback. This phenomenon is also one of the central points of this research. Zhan's model, presented in the previous section, is incapable of modelling the creep strain recovery after unloading from creep because it only deals with elastic springback. The purpose of the present work, therefore, is to develop a new set of constitutive equations, while retaining the major virtue of Zhan's model, to overcome the above limitations and predict anelastic creep recovery after creep age forming.

5.3.1 Introduction

The development of the new constitutive model, generally speaking, has passed through three levels of increasing complexity for different purposes. The fundamental level refers to the original model proposed by Zhan for the predictions of solid solution strengthening, dislocation hardening, and age hardening of the alloy from an initial stage of solution heat treatment, or the T3/T4 condition. It is later extended to model more sufficiently the softening behaviour of the alloy after the peak-aged condition, or the T6/T7 condition. In the third level, through the development of a new set of equations, the phenomenon of anelastic creep recovery will be covered. The new constitutive model, similar to other constitutive models, requires fitting of the experimental data for the determination of material constants.

The basic strategies of developing the new constitutive equations can be summarised below. All of these strategies have been applied for the development of a new set of equations.

- Formulating new evolution equations with state variables, based on the theories and experimental data, to represent the desired new micro-mechanisms,
- Creating and adding more adjustable material constants to the constitutive equations for the purpose of fitting more flexibly to the experimental data,
- Modifying the form of the evolution equations, while keeping the adjustable material constants the same, to describe more adequately the micro-mechanisms concerned, and
- Removing the redundant adjustable material constants and functions of the evolution equations, while not influencing the characteristics of state variables representing the

micro-mechanisms concerned, so as to achieve a simpler form for further numerical implementations.

The next section will illustrate the steps to develop the new set of constitutive equations: (i) to explore the micro-mechanisms and phenomena covered in this model; (ii) to discuss the key assumptions and theories behind the equations; (iii) to formulate the required equations to cover the micro-mechanisms of time-dependent inelastic and anelastic deformation; and (iv) to detail the development of the new set of constitutive equations. The new set of equations will finally be expanded into a three-dimensional form for further numerical integrations. The numerical integrations of this new set of equations into FE simulations will be discussed in the next chapter.

5.3.2 The mechanisms and phenomenological bases behind anelasticity

As reviewed in Chapter 2, a number of micro-mechanisms are involved in anelastic creep recovery. They can be generalised into two main categories: the shearing and sliding of grain boundaries, and the formation and dissipation of dislocation cell structures.

While various theoretical treatments of the mechanisms regarding grain boundary sliding have been proposed [1, 2, 177], to the author's knowledge, none have been shown to have quantitative accuracy sufficient to predict the shearing behaviour and the energy stored in grain boundaries that cause anelastic creep recovery. On the other hand, the micro-mechanisms of dislocation motion have become a generally approved concept – the state variable of normalised dislocation density ($\bar{\rho}$) has been used in Zhan's model and many other models [48, 178, 179] – accounting for an intrinsic strengthening effect due to the material's deformation in the forming process.

The approach taken in the development of a new set of equations was therefore to choose the suitable mechanisms with regards to the formation and dissipation of dislocation cell structures – a phenomenologically driven form of generation, movement, and dissipation of mobile and immobile dislocations and their corresponding evolutions – for the expression of the creep recovery behaviour.

In Chapter 4, we have determined the viscous and anelastic creep strain by creep aging and recovery test (CART), as shown in Figure 4.20. The key assumptions are that mobile

dislocations play a role in affecting the anelastic strain and that immobile dislocations play a role in affecting the viscous strain.

Mobile dislocations themselves possess long-range internal stress fields. During primary creep, mobile dislocations gradually develop and increase with time. Some of those dislocations gradually become entangled with each other or pinned by precipitates and thus immobilised. The rest of the mobile dislocations move into low energy configurations, but their long-range internal stress fields are not neutralised [124, 180, 181]. These internal stress fields are accommodated in the opposite direction to the applied stress in the material. An illustration of the evolutions of microstructures during the creep age forming process is presented in Figure 5.4. The process can be described in several steps as follows:

- In the initial state, a metallic material with no dislocations in the matrix is considered. Depending on the heat treatment conditions, some or no precipitates may be present in the matrix.
- When primary creep occurs, several mechanisms are involved, including dislocation bowing between obstacles, dislocation pileups, and the formation of dislocation cell structures. Dislocations participating in these mechanisms are treated as mobile dislocations. These mobile dislocations are gradually generated, and some of the mobile dislocations transform into immobile dislocations as they become entangled with each other and are pinned by obstacles. The remaining mobile dislocations gradually move into low energy configurations.
- In secondary creep, as the mobile dislocation density reaches a saturation level, the long-range elastic stress fields of mobile dislocations are also balanced by their low energy formations – dislocation cell structures. In this state, the number of immobile dislocations is still increasing.
- When the material is unloaded from creep, the change in internal stress causes low energy configurations to break down, and the dislocation cell structures gradually dissipate.

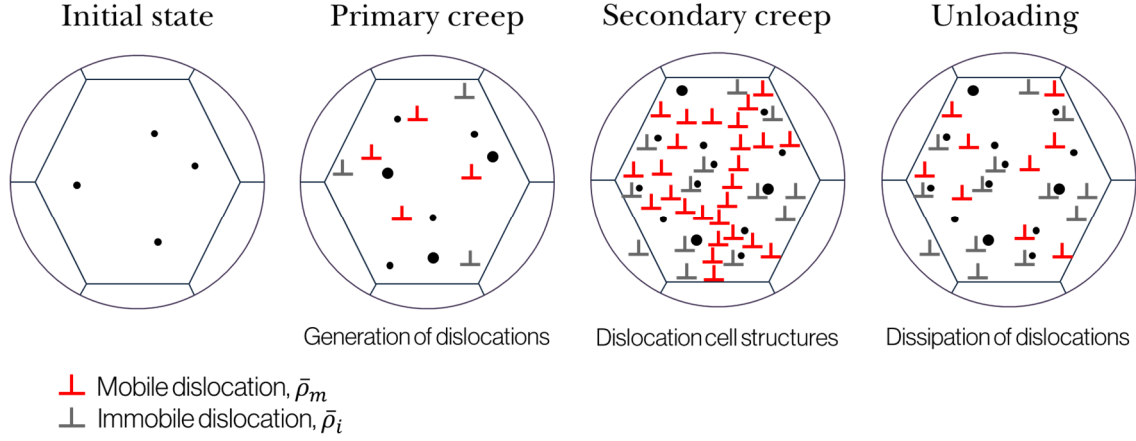


Figure 5.4. Schematic diagrams of microstructural variations in different stages of creep aging.

5.3.3 Formulation of the equations

The equations are formulated based on the synthesis of physical mechanisms discussed in the previous sections and experimental results discussed in Chapter 4. The mathematic framework can be categorised as the equations relating to internal stresses and creep strain, and the ones relating to hardening variables.

5.3.3.1 The role of back stresses on the creep strain

With all of the mechanisms stated in the previous section, back stresses can be seen as the driving force that propels anelastic creep recovery. For dislocation bowing, the energy generates extra-long dislocations, and the tension of dislocation lines creates an internal force acting in the opposite direction of the force, which straightens the bowing segment of the dislocation [182]. The pileup of nearby similar dislocations creates a stress field acting on a leading dislocation, which is opposite to the direction of the motion of the dislocations [183]. Dislocation cell structures formed in secondary creep have a very important property: Under an applied stress, σ , the interior of the cell wall possesses long-range internal stresses acting oppositely to the applied stress.

As all of the above mechanisms are involved in the resulting directional internal stresses, it is convenient, in a mathematical manner, to use back stress, σ_b , to represent the net effect of those internal stresses for incorporation into the new model to describe the material's response to unloading.

For creep deformation, the creep strain rate, $\dot{\epsilon}_c$, can be seen as being affected by an effective stress – the superposition of the applied stress, σ , and back stress, σ_b :

$$\dot{\epsilon}_c = f(|\sigma - \sigma_b|) \quad (5.10)$$

Based on the experimental results from CABTs, as presented in Chapter 4.3.3, it is appropriate to presume that the back stress, σ_b , is proportional to the applied stress σ at the primary creep state, as shown in Figure 4.23. Since mobile dislocations are assumed to be the main cause of back stress, the relationship between back stress, σ_b , and mobile dislocation density, ρ_m , can be written as below:

$$\dot{\sigma}_b \propto \bar{\rho}_m^{n_x-1} \cdot \dot{\bar{\rho}}_m \quad (5.11)$$

where $\bar{\rho}_m$ is the normalised mobile dislocation density normalised by equation (5.9) and n_x is a constant. When the material is unloaded from creep,

$$\sigma = 0 \quad (5.12)$$

The internal stress caused by mobile dislocations generates a backflow in the material that moves the mobile dislocation backwards and causes anelastic creep recovery. At this stage, by substituting equation (5.12) into (5.10), the backflow behaviour of the creep strain after unloading can be rewritten as follows:

$$\dot{\epsilon} = f(|-\sigma_b|) \quad (5.13)$$

The evolution rate of the normalised mobile dislocation density is assumed to be a function of itself and creep strain rate ($\dot{\epsilon}_c$), as shown in equation (5.14) below:

$$\dot{\bar{\rho}}_m \propto f_1(\bar{\rho}_m, \dot{\epsilon}_c) - f_2(\bar{\rho}_m) \quad (5.14)$$

The first term of the equation above represents the behaviour that, in the primary creep stage, mobile dislocations are proportional to the creep strain rate and increase with time. The second term represents the static recovery process in the secondary creep stage. At that stage, the annihilation of oppositely-signed dislocations and the formation and coarsening of dislocation cell wall structures cause the total amount of mobile dislocations to stop increasing and reach a saturated level.

5.3.3.2 The hardening variables

In section 4.2, microstructural investigation has been performed to elucidate the strengthening mechanisms. There is, however, lack of available microstructural data to quantify hardening effect. The present model, therefore, follow Zhan's model, where the yield strength of the material is contributed by three internal state variables: solution hardening, dislocation hardening, and age hardening variables, as shown in equation (5.15). Each state variable will be discussed in detail in the upcoming sections.

$$\sigma_y = \sigma_{ss} + \sqrt{\sigma_a^2 + \sigma_{dis}^2} \quad (5.15)$$

a. Dislocation hardening

The immobile dislocations contribute to permanent strain on the material. Their contribution to the yield strength follows the Taylor hardening relationship, $\sigma_y|_{\epsilon, T} \propto E\sqrt{\rho}$ [184]. In contrast, mobile dislocations are not associated with strain hardening because they possess long-range elastic stress fields that dissipate after unloading from creep. The contribution of immobile dislocation densities to the yield strength can be written as follows:

$$\dot{\sigma}_{dis} \propto f(n \cdot \bar{\rho}_i^{n-1}) \quad (5.16)$$

where $\bar{\rho}_i$ is the normalised immobile dislocation density. It is normalised by using equation (5.9), and n is a material constant. The formation of immobile dislocations is assumed to arise from the pinning and entangling of mobile dislocations, and it is driven by the current magnitude of the creep strain rate $\dot{\epsilon}_c$. The generation of immobile dislocations is also influenced by the dynamic recovery of immobile dislocations, which decreases the immobile dislocation density. Combining those effects, the evolution rate of immobile dislocation density can be written as

$$\dot{\bar{\rho}}_i = f_1(\bar{\rho}_m, \epsilon_c) - f_2(\sigma_y, \bar{\rho}_i) \quad (5.17)$$

b. Age hardening

In Figure 4.11, the longitudinal length of GP zones and η' precipitates under pure aged and creep aged conditions has been measured. How these data can relate to age hardening, however, are still inconclusive. The evolution of precipitate radius, therefore, is treated as

internal state variables that are determined by the model but not from the experimental data. The evolution rate of precipitates can be defined as

$$\dot{\bar{r}} = C_r(Q - \bar{r})^{m_3} \quad (5.18)$$

where $\bar{r}$ is the normalised precipitate radius, normalised by equation (5.8), and Q and m_3 are material constants. This equation, unlike Zhan's equations, ignores the influence of dislocations on the precipitate evolution rate. The reason is that the strength of alloy AA7050 results mainly from contributions by GP zones and η' precipitates that are coherent or semi-coherent to the matrix. The strength contributed by precipitates follows Zhan's equation:

$$\dot{\sigma}_a = C_A \dot{\bar{r}}^{m_1} (1 - \bar{r}) \quad (5.19)$$

c. Internal strength

The definition of internal strength is the strength related to solid solution strengthening, grain boundary hardening, and intrinsic hardening of the matrix [185]. The temper condition of the alloy in the present study, AA7050-TAF, is the peak aged condition. It is assumed that most of the alloying elements in the matrix have already been incorporated into precipitates. The solid solution evolution rate, therefore, can be treated as zero:

$$\dot{\sigma}_{ss} = 0 \quad (5.20)$$

In grain boundary hardening, the motion of dislocations can be impeded by grain boundaries. Although grain size influences yield strength, it is assumed that it remains unchanged during creep age forming because of the low deformation in the process. The strength contributed by grain boundaries can therefore be seen as a constant. Intrinsic strength is defined as the strength of the aluminium matrix, which is also a constant. We can use a constant to represent the strength contributed by this strengthening effect:

$$\sigma_{ss} = \text{constant} \quad (5.21)$$

5.3.4 Final Equations for the model

Based on the analysis in the previous chapters for the development of the constitutive equations with the consideration of anelastic creep recovery due to the backflow of internal stress, or back stress, the final uniaxial constitutive equations can be formulated as follows:

$$\dot{\epsilon}_c = A_1 \sinh\{B_1[(\sigma - n_0\sigma_b) \cdot (1 - \bar{\rho}_m)]\} \quad (5.22)$$

$$\dot{\bar{\rho}}_m = A_3(1 - |\bar{\rho}_m|)(\dot{\epsilon}_c) - C_p|\bar{\rho}_m|^{m_5} \cdot \text{sign}(\bar{\rho}_m) \quad (5.23)$$

$$\dot{\sigma}_b = D_2 \cdot n_x \cdot \bar{\rho}_m^{n_x-1} \cdot \dot{\bar{\rho}}_m \quad (5.24)$$

$$\dot{r} = C_r(Q - \bar{r})^{m_3} \cdot (1 + r_0\bar{\rho}_m^{m_4}) \quad (5.25)$$

$$\dot{\sigma}_a = C_A \dot{r}^{m_1}(1 - \bar{r}) \quad (5.26)$$

$$\sigma_y = \sigma_{ss} + \sqrt{\sigma_a^2 + \sigma_{dis}^2} \quad (5.27)$$

$$\dot{\bar{\rho}}_i = A_4\bar{\rho}_m\epsilon_c - A_5\sigma_y\bar{\rho}_i^{\frac{3}{2}} \quad (5.28)$$

$$\dot{\sigma}_{dis} = A_2 \cdot n \cdot \bar{\rho}_i^{n-1} \dot{\bar{\rho}}_i \quad (5.29)$$

This is a complete set of equations characterising creep aging and anelastic creep recovery behaviour in the creep age forming process. The term used in equation (5.22) to describe the evolution of the creep strain rate $\dot{\epsilon}_c$ is the effective stress $(\sigma - n_0\sigma_b)$ with a usual hyperbolic sine function. $\dot{\epsilon}_c$ is positive when the material is being deformed under a certain stress level in the creep age forming process and becomes negative when the material is unloaded.

As previously stated, both the normalised mobile dislocation density $\bar{\rho}_m$ and the normalised immobile dislocation density $\bar{\rho}_i$ are introduced as the state variables for evolution equations. The fundamental difference between them is that $\bar{\rho}_m$ contributes back stress to the material and dissipates with time in unloading, whereas $\bar{\rho}_i$ is defined as the dislocations that cause permanent deformation in the material and do not contribute any internal stress to the material. Since, in effect, some of the $\bar{\rho}_m$ gradually transform into $\bar{\rho}_i$, it can be concluded that a proper form of the equation for the generation of immobile dislocation densities is (5.28).

The creep strain rate $\dot{\epsilon}_c$ has a substantial influence on the evolution rate of mobile dislocations $\dot{\bar{\rho}}_m$, since mobile dislocations increase with time in primary creep and reach a saturation state in secondary creep due to the formation of dislocation cell structures for their low energy configurations.

When $\dot{\epsilon}_c$ is reversed due to unloading, $\dot{\bar{\rho}}_m$ also becomes negative because mobile dislocations gradually dissipate due to the breakdown of dislocation cell structures. This relationship is characterised in equation (5.23). It should be noted that an absolute value function is placed on $\bar{\rho}_m$ to stand for the fact that the number of mobile dislocation densities cannot become zero.

As the back stress, σ_b , is defined as the net effect of the internal stresses caused by mobile dislocations, the amount of $\bar{\rho}_m$ has a direct influence on $\dot{\sigma}_b$. Equation (5.24) delineates the relationship between $\dot{\sigma}_b$ and $\bar{\rho}_m$.

5.3.5 Determination and validation of material constants

The new constitutive equations consider creep deformation, anelastic creep strain recovery, back stress, state variables related to microstructure evolutions, and hardening variables. Thus, only conducting more experiments with CART and CABT is insufficient for the determination of material constants. Substantially, more material constants than in Zhan's model, 19 constants in total, need to be determined.

These material constants can be obtained in a systematic way based on numerical and computational interpretations of those constants in relation to the phenomena and micro-mechanisms of the material. This approach involves separately identifying creep strain, recovery strain, microstructure variables, hardening variables, and yield strength. The determination process of the material constants is similar to the one demonstrated in Section 5.2.2.

A combination of CART and CABT coupled with accurate strain measurement, and interrupted uniaxial tensile tests after creep aging tests were used to determine the constants for the new model. The experimental technique has been presented in Chapter 3.4 and the experimental results, in Chapter 4.3. The initial values were determined first based on the understanding of the current material, as shown in Table 5.3 and discussed in

Section 5.2.1. The material constants for constitutive equations determined from the method above are listed in Table 5.4.

Table 5.3. Initial value of the new constitutive model for creep and anelastic recovery.

ε_c	$\bar{\rho}_m$	σ_b (MPa)	$\bar{r}$	σ_a (MPa)	σ_{ss} (MPa)	$\bar{\rho}_i$	σ_{dis} (MPa)
0	0	0	1	300	273	0	0

Table 5.4. Constants of the new constitutive equations for modelling the anelastic behaviour of the AA7050-TAF alloy at 174°C.

A_1	B_1	n_0	A_3	C_p	m_5	D_2
9.87e-06	0.054	6.6	316	2988	6	124.8
n_x	C_r	Q	m_3	m_4	γ_0	C_a
1.07	0.037	5	1.3	2.4	40.7	80.2
m_1	A_4	A_5	A_2	n		
0.97	50	0.001	10	0.1		

Figure 5.5 compares the experimental results from CART and the model fit. The elastic strains caused by loading and unloading have the same value and are excluded from data for plots of strain versus time in Figure 5.6. As seen in the plots, anelastic creep strain recovery can be modelled when the applied stress is removed. Figure 5.5a presents the results of the tests under different applied stresses, 137.5, 150, and 162.5 MPa, for the same creep aging time. Figure 5.5b displays the trend of anelastic creep recovery after creep aging for 2, 4, 6, and 8h under 150 MPa. Good agreement can be obtained on the creep aging part, and a fair result with maximum error of 7% is achieved in the recovery part.

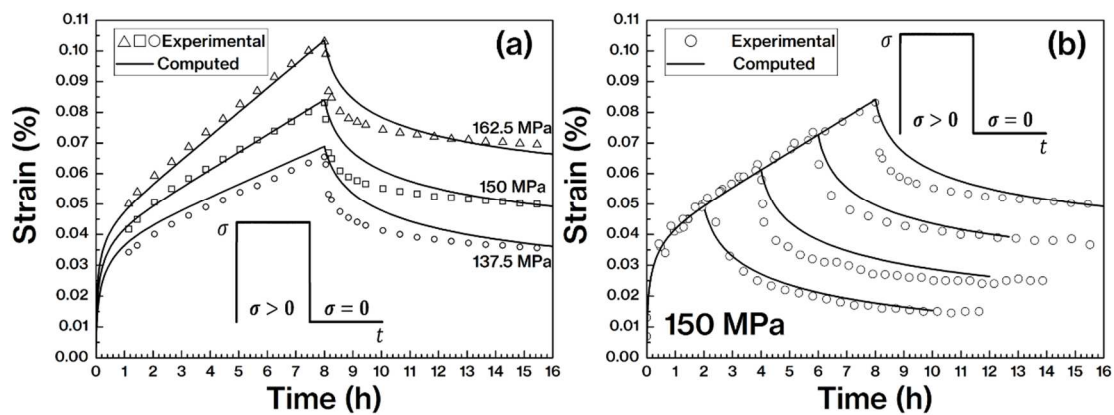


Figure 5.5. Experimental (symbols) and computed (solid lines) curves of creep aging and recovery tests (CART) (a) for 8h under 137.5, 150, and 162.5 MPa, and (b) for 2, 4, 6, and 8h under 150 MPa respectively.

Figure 5.6a shows the predicted normalised precipitate radius evolution computed with the new model. Since the growth of precipitates is strongly influenced by the aging temperature but not so sensitive to the applied stress, the precipitates continue to grow after unloading. In contrast, the mobile dislocations, as shown in Figure 5.6b, dissipate after unloading and cause anelastic creep recovery. Immobile dislocations stop growing after unloading and remain unchanged afterwards. The comparison of computed yield strength evolutions data and the experimental yield strength data from Figure 4.4 are presented in Figure 5.7a. The results are similar to the predictions by Zhan's equations, as shown in Figure 5.2. Figure 5.7b shows the fit of the back stress model with the experimental results from the CABT. Since the back stress is controlled by the mobile dislocation density, as shown in Figure 5.6a, their evolutionary trends are very similar.

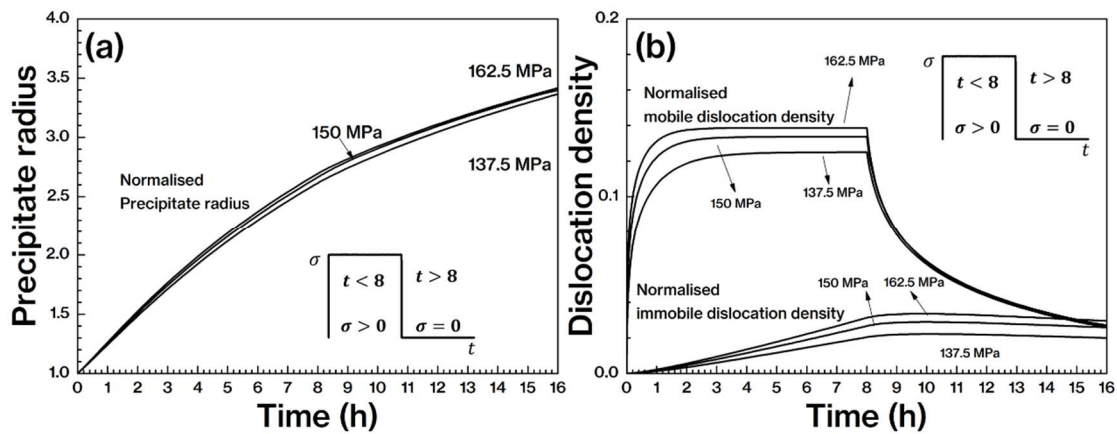


Figure 5.6. (a) Computed normalised precipitate size variation (b) Computed mobile and immobile dislocation density variation for the new equation.

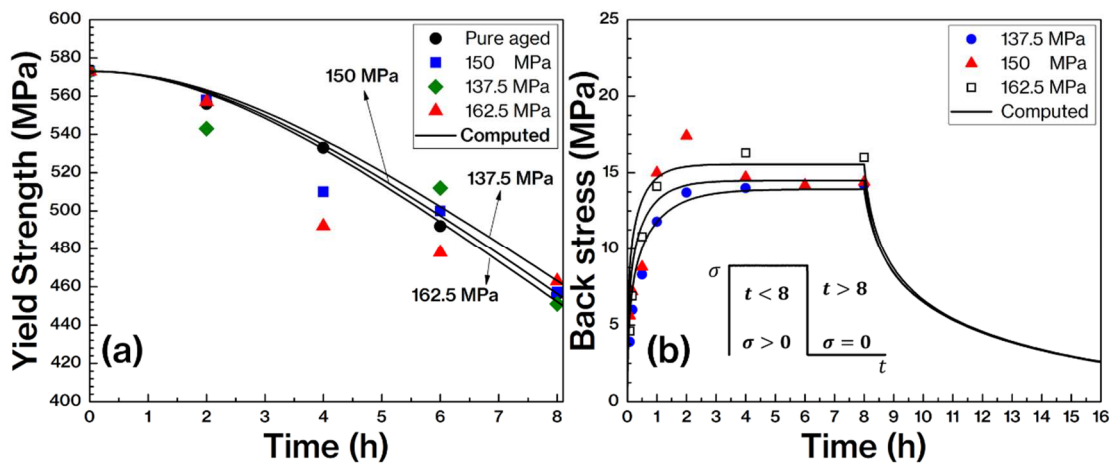


Figure 5.7. Experimental (symbols) and computed (solid lines) (a) yield strength evolution under different stress levels, and (b) back stress evolution under different stress levels for 8h, computed with the new equation.

5.3.6 Equations for anelasticity under multiaxial stresses

In the preceding sections, efforts have been made to develop the uniaxial version of the equations. All of these concepts can be generalised, through very direct assumptions, to a full multiaxial set of equations. The multiaxial model contains a mixture of isotropic and kinematic hardening that simulates yield surface expansion and translation, where the creep deformation occurs when the elastic energy of distortion reaches a critical value. The implementation can be specified as follows:

- The applied stress σ and back stress σ_b become tensor forms σ_{ij} and σ_{ij}^b . An effective stress tensor is defined by $\sigma_{ij}^{eff} = \sigma_{ij} - n_0 \sigma_{ij}^b$.
- The effective stress tensor σ_{ij}^{eff} for creep deformation and anelastic creep recovery is then replaced by an equivalent von Mises stress $\bar{\sigma}^{eff} = \sqrt{\frac{3}{2} s_{ij} : s_{ij}}$, where s_{ij} are components of the stress deviator tensor, $s_{ij} = \sigma_{ij}^{eff} - \frac{1}{3} \sigma_{kk}^{eff} \delta_{ij}$.
- Similar to $\bar{\sigma}^{eff}$, a von Mises-like effective value of mobile dislocation density $\bar{\rho}_m^{eff}$ is used to represent the quantity of mobile dislocations in any direction, where $\bar{\rho}_m^{eff} = \sqrt{\frac{3}{2} T_{ij} : T_{ij}}$, and $T_{ij} = \bar{\rho}_{ij}^m - \frac{1}{3} |\bar{\rho}_{kk}^m| \delta_{ij} \cdot \text{sign}(\bar{\rho}_{ij}^m)$.
- For the evolution rate of mobile dislocation densities $\dot{\bar{\rho}}_{ij}^m$, instead of using $\dot{\epsilon}^c$, the evolution equation uses $(1 - |\bar{\rho}_{ij}^m|)(\dot{\epsilon}_{ij}^c)$ in the first term with a statement of (i,j) no sum, where i,j is for the individual components. This is to follow the uniaxial case but has its own components so as to show the influence of each component of the strain rate on mobile dislocation densities. A sign function is added to the equation to delineate that the second term always has the opposite sign of the first term. Specifically,

$$\dot{\bar{\rho}}_{ij}^m = A_3 (1 - |\bar{\rho}_{ij}^m|)(\dot{\epsilon}_{ij}^c) - C_p |\bar{\rho}_{ij}^m|^{m_5} \cdot \text{sign}(\bar{\rho}_{ij}^m), (i,j) \text{ no sum} \quad (5.30)$$

- The evolution rate of the back stress component $\dot{\sigma}_{ij}^b$, similar to $\dot{\bar{\rho}}_{ij}^m$, also uses a statement of (i,j) no sum to represent the influence of each component of the $\dot{\bar{\rho}}_{ij}^m$ on $\dot{\sigma}_{ij}^b$.
- The effective multiaxial strain rate $\dot{\epsilon}^c$ is introduced for creep and anelastic creep recovery based on the assumption of the normality of flow rule, i.e.,

$$\dot{\bar{\epsilon}}_{ij}^c = \frac{3}{2} \frac{s_{ij}}{\bar{\sigma}^{eff}} \cdot \dot{\bar{\epsilon}}^c \quad (5.31)$$

- All the remaining hardening variables, $\bar{r}$, $\bar{\rho}_l$, σ_a , and σ_{dis} , are assumed to be isotropic and thus remain in the same form as the uniaxial equations.

Following these principles, a multiaxial generalisation has been made on the uniaxial version of the equations. The full set of multiaxial unified creep and anelastic constitutive equations can be stated as follows:

$$\dot{\bar{\epsilon}}^c = A_1 \sinh\{B_1[(\bar{\sigma}^{eff}) \cdot (1 - \bar{\rho}_m^{eff})]\} \quad (5.32)$$

$$\sigma_{ij}^{eff} = \sigma_{ij} - n_0 \sigma_{ij}^b \quad (5.33)$$

$$\dot{\bar{\epsilon}}_{ij}^c = \frac{3}{2} \frac{s_{ij}}{\bar{\sigma}^{eff}} \cdot \dot{\bar{\epsilon}}^c \quad (5.34)$$

$$\bar{\sigma}^{eff} = \sqrt{\frac{3}{2} s_{ij} : s_{ij}} \quad (5.35)$$

$$s_{ij} = \sigma_{ij}^{eff} - \frac{1}{3} \sigma_{kk}^{eff} \delta_{ij} \quad (5.36)$$

$$\dot{\bar{\rho}}_{ij}^m = A_3(1 - |\bar{\rho}_{ij}^m|)(\dot{\bar{\epsilon}}_{ij}^c) - C_p |\bar{\rho}_{ij}^m|^{m_5} \cdot \text{sign}(\bar{\rho}_{ij}^m), (i, j) \text{ no sum} \quad (5.37)$$

$$\dot{\sigma}_{ij}^b = D_2 \cdot n_x \cdot |\bar{\rho}_{ij}^m|^{(n_x-1)} \cdot \dot{\bar{\rho}}_{ij}^m, (i, j) \text{ no sum} \quad (5.38)$$

$$\bar{\rho}_m^{eff} = \sqrt{\frac{3}{2} T_{ij} : T_{ij}} \quad (5.39)$$

$$T_{ij} = \bar{\rho}_{ij}^m - \frac{1}{3} |\bar{\rho}_{kk}^m| \delta_{ij} \cdot \text{sign}(\bar{\rho}_{ij}^m) \quad (5.40)$$

$$\dot{\bar{r}} = C_r(Q - \bar{r})^{m_3} \cdot (1 + r_0(\bar{\rho}_m^{eff})^{m_4}) \quad (5.41)$$

$$\dot{\sigma}_a = C_A \dot{\bar{r}}^{m_1} (1 - \bar{r}) \quad (5.42)$$

$$\dot{\bar{\rho}}_l = A_4 \bar{\rho}_m^{eff} \bar{\epsilon}^c - A_5 \sigma_y \bar{\rho}_l^{\frac{3}{2}} \quad (5.43)$$

$$\dot{\sigma}_{dis} = A_2 \cdot n \cdot \bar{\rho}_l^{n-1} \dot{\bar{\rho}}_l \quad (5.44)$$

5.3.7 Summary

The development and capabilities of the new set of constitutive equations for anelastic creep recovery have been demonstrated in detail. These equations, like most other unified constitutive equations, use a single variable of non-elastic strain to predict both creep and anelasticity. Furthermore, by using a ‘back stress’ variable and other evolution equations to predict the rate of change of the internal stresses and other structure variables, the equation can model complex micro-mechanisms and hardening effects in a rational manner.

A multiaxial version of the model has been developed. The next chapter will demonstrate a numerical method for integrating the multiaxial equations into the commercial finite element software ABAQUS for further structural analysis.

6 MODEL VALIDATION AND PREDICTION

This chapter demonstrates the implementation and use of the constitutive equations in ABAQUS, a commercial FE software. An integrated numerical procedure to simulate the creep age forming process is also presented. The stress relaxation behaviour and the effect of anelasticity on springback are analysed through FE simulations. The experimental results of three-point bending stress relaxation tests presented in Chapter 3 are used to validate the simulation.

6.1 Introduction

Quantitative predictions of a material's behaviour in the metal forming process are of enormous interest in manufacturing industries. Although forming tests are inevitable, an accurate constitutive model accompanied by FE simulation would allow a considerable reduction in the number of forming trials to meet the expected qualities.

The work presented in the previous chapter has demonstrated that it is possible, through constitutive equations, to model the material's deformation behaviour during the creep age forming process and to model the effect of anelastic creep recovery after unloading. The proposed model considers microstructural variations. Many of the microstructural features and mechanical properties, such as the evolutions of mobile/immobile dislocation densities, precipitates, and yield strength, have been delineated in the model.

To be able to predict and describe inelastic/anelastic behaviour and micro-mechanisms of materials through constitutive modelling is the key to modelling; for industrial applications, however, people are more concerned about the application of these constitutive equations to predict the formed parts in the actual forming process, where more complex shapes are involved.

With the aim of providing a practical technique to make quantitative predictions of the creep age forming process, an integrated numerical procedure to simulate the creep age forming process has been developed. The simulation is mainly concerned with evaluating the effect of anelastic creep recovery on springback after the creep age forming process. In order to evaluate the predictive capability of the constitutive equations, FE model predictions of the three-point bending stress relaxation and back stress test are compared with the experimental results presented in chapter 3.

6.2 Description of the FE model

The FE simulations of the creep age forming process were conducted using the commercial software ABAQUS. The setup and steps of the simulations followed the three-point bending stress relaxation and back stress tests presented in Chapter 3. A two-dimensional finite element model was built to represent the rollers and the AA7050-TAF alloy plate, as shown in Figure 6.1. The boundary and loading conditions will be explained later.

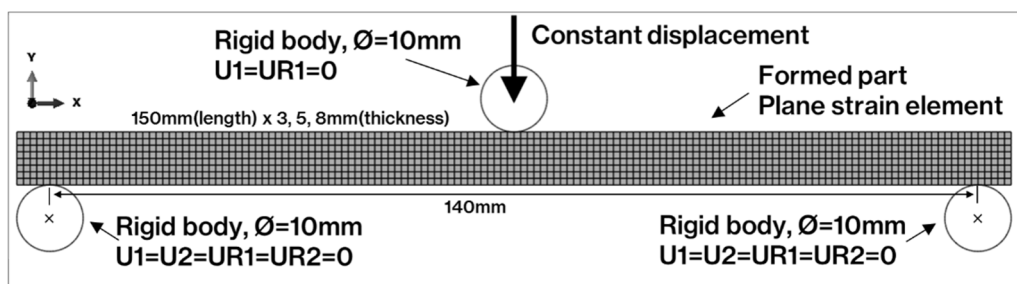


Figure 6.1. Boundary and loading conditions of the finite element model.

The simulations were performed on aluminium plate of three thicknesses: 3, 5, and 8mm, respectively. The rollers were defined as rigid bodies of 10mm diameter, and the length of the aluminium plate was 150mm. The friction coefficient was set to 0.2 at the contact surface of the roller and the aluminium plate to simulate non-lubricated contact surfaces. The rollers were meshed by using an analytical rigid model with an element size of 0.5mm.

For the aluminium plate, 4-node two-dimensional quadrilateral element (CPE4R) plane strain elements were used for the simulations.

In order to analyse the stress distribution along the thickness during the simulations, finer meshes were used along the thickness. The element sizes (thickness x length) were set at 0.3 x 1, 0.5 x 2.5, and 1 x 1mm respectively for 3, 5, and 8mm thick plate. Thermo-mechanical behaviour is not considered in this model because all the SRBT experiments and FE simulations are performed in isothermal condition (174°C), where the rollers and the samples are at the same temperature throughout the process. The material's creep aging properties at that temperature are defined inside the constitutive model.

6.3 Numerical Procedures

6.3.1 Material model implementation

The set of multi-axial creep-aging constitutive equations presented in the previous chapter was implemented into the ABAQUS finite element code by means of the user defined subroutine VUMAT. The VUMAT subroutine, written in Fortran, uses explicit differential integration (the Euler explicit method) to update the vector-formed stress and state variables in the algorithm [186]. The code of the subroutine is presented in Appendix B.

6.3.2 Overall simulation process

Following the experimental procedure of the three-point bending stress relaxation and back stress (SRBT) test presented in Chapter 3, the simulation process consisted of five distinctive steps to simulate stress relaxation behaviour and anelastic creep strain and load recovery behaviour after stress relaxation. All of the steps in the simulations were performed in isothermal conditions, 174°C.

As illustrated in Figure 6.2, two kinds of simulations, namely, the Strain Recovery (SR) and Load Recovery (LR) cases, were performed to evaluate the effects of creep strain recovery and total load recovery on the springback. The modelling procedures for both SR and LR were the same in step 1 and step 2. They differed from step 3 to step 5.

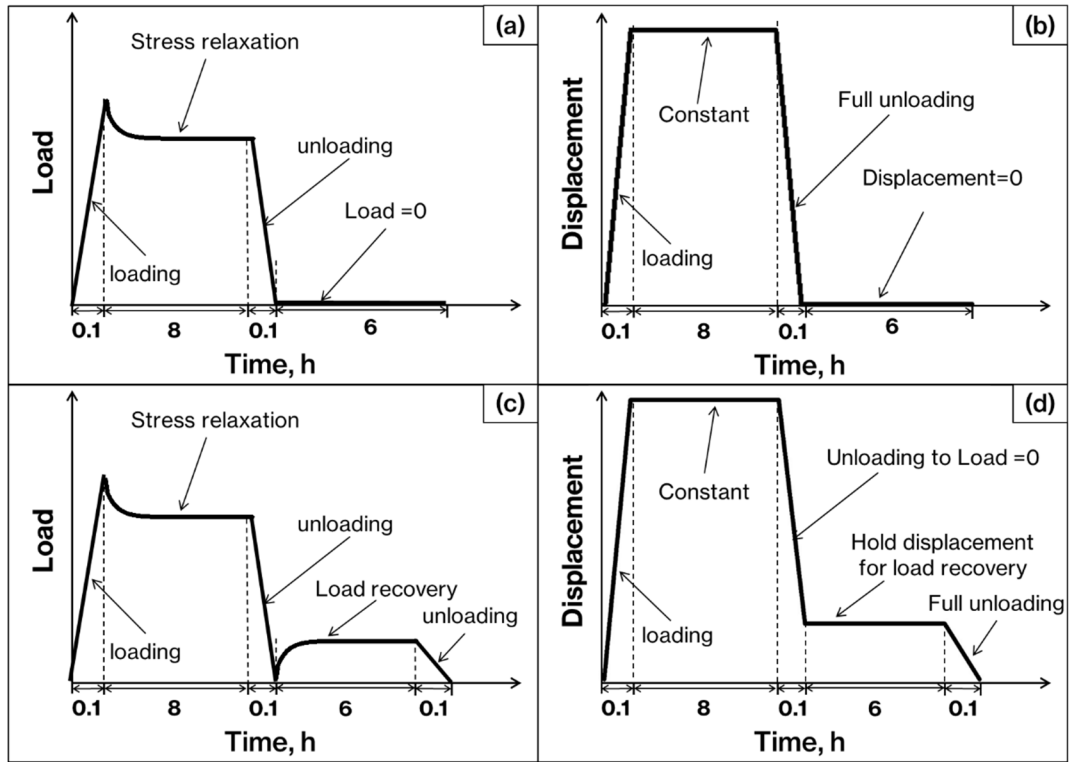


Figure 6.2. Schematic diagram of the loading and displacement configuration of Strain recovery (SR) simulations, (a) and (b), and Load recovery (LR) simulations, (c) and (d).

The steps for the simulation can be summarised as follows:

- **Boundary condition:** A boundary condition of $U1=U2=UR1=UR2=0$ was set for the lower parts of the two rollers, and $U1=UR1=0$ was given to the upper roller. Thus, there were no translational (U) or rotational (UR) movements along the respective directions ($1 = x$, $2 = y$). These boundary conditions were maintained throughout the entire simulation process.
- **Step 1 – Elastic loading:** A concentrated force was applied onto the upper roller using the displacement control method to deflect the plate. This step ended when the displacement, $U2$, of the upper roller reached the desired value. The duration of this step was 0.1h, with a fixed time increment of $1e-7$ h.
- **Step 2 – Stress relaxation:** The deflection of the plate was maintained throughout this step by holding the displacement of the upper roller, $U2=Constant$, to enable stress relaxation to occur in the plate. The duration of this step was 8h, with a fixed time increment of $5e-8$ h.

For the Strain recovery (SR) case:

- **Step 3 – Elastic unloading:** The concentrated force applied on the upper roller was fully removed with the displacement control method by raising the displacement of the upper roller to the original point, $U_2=0$. Instant springback of that point could be measured at the end of this step. The duration of this step was 0.1h, with a fixed time increment of 5e-8h.
- **Step 4 – Strain recovery (SR):** The plate was “load free” in this step. The plate would gradually return towards the original shape due to anelastic creep recovery. The duration of this step was 6h, with a fixed time increment of 5e-8h.

For the Load recovery (LR) case:

- **Step 3 – Elastic unloading:** The concentrated force applied on the upper roller was removed by gradually raising the upper roller. The upper roller stopped at the point where the concentrated force of the upper roller reached zero. Instant springback of that point could be measured at the end of this step. The duration of this step was 0.1h, with a fixed time increment of 5e-8h.
- **Step 4 – Load recovery (LR):** Similar to step 2, the displacement of the upper roller was maintained, $U_2=\text{Constant}$, to hold the deflection of the plate. Since the anelastic creep strain recovery could not take place, force generated on the plate, acting against the roller, could be expected. The duration of this step was 6h, with a fixed time increment of 5e-8h in both cases.
- **Step 5 – Full unloading:** In this step, the roller moved back to the original point, $U_2=0$, to remove the load generated in step 4. The duration of this step was 0.1h, with a fixed time increment of 1e-7h .

6.4 Model validation

In order to validate the implementation of the subroutine, a three-dimensional single element simulation was performed first to verify the accuracy of finite element solutions. The single element model was constructed as a cubic shape using a 3D stress element

(C3D8R) with a global size of 1mm. Young's modulus $E=65$ GPa and Poisson's ratio $\nu=0.32$ were defined for AA7050-TAF alloy in the simulation.

The process of the single element simulation was simply to fix one end of the cubic element with the boundary condition of $U1=U2=U3=UR1=UR2=UR3=0$ and apply a uniformly distributed stress using the load control method with a loading condition of $U1=137.5, 150$, or 162.5 MPa on to the other end of the cubic element. The stress was then released after 8h of creep aging for the anelastic creep recovery step.

Figure 1.1 presents a comparison of the computed creep aging strain-time curves acquired from the numerical solution by Matlab using the Runge Kutta method with the FE simulation results from ABAQUS. Good agreement of the numerical and the FE simulation results can be observed.

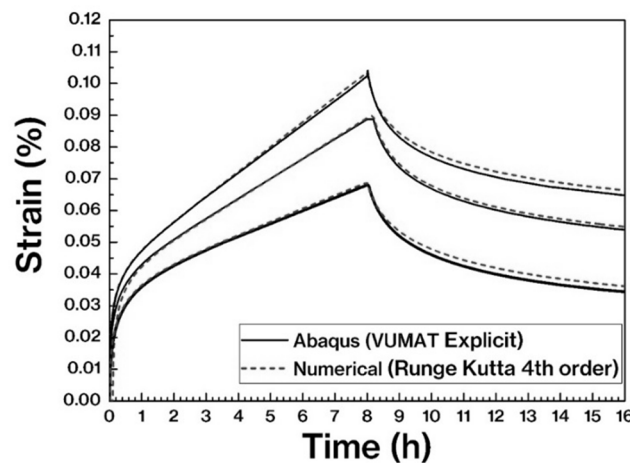


Figure 6.3. Comparisons of the creep aging strain-time curves of the numerical solution, Runge Kutta 4th order (dashed lines), and ABAQUS/Explicit (solid line).

Mesh convergence studies were performed to ensure that the results of the simulations were not affected by changing the size of the mesh. Figure 6.4a shows the time-dependent solution of load recovery behaviour of SRBT simulation for plate of 5mm thickness and 4mm deflection. The results were obtained with different numbers of elements, i.e., 150, 300, 600, and 1,200 meshes. The load values of these curves acquired at 8h are plotted against time in Figure 6.4b. Good agreement can be observed for the 600 and 1,200 meshes. Considering time and efficiency of computation, the solution with 600 meshes was chosen, with an aspect ratio of element of 1:2.5.

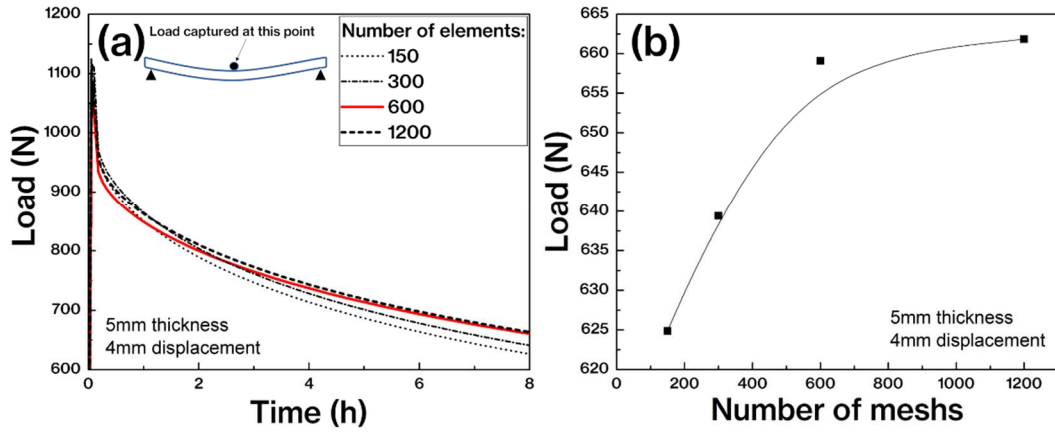


Figure 6.4. Convergence of time-dependent solution for the SRBT simulation of 5mm thickness and 4mm displacement showing (a) the influence of the number of meshes on the accuracy of the solution, (b) a typical 4-point convergence curve showing the load collected at 8h of SRBT simulation.

The time increments of each iteration are also important in these types of analysis because the result can only be accurate when the increments are sufficiently small. In order to determine the ideal size of the time increment, three-point bending stress relaxation simulations with different time increments were performed first. The size of the increment was gradually reduced in these simulations. From the results of trial simulations, an optimum value for the stable time increment was chosen.

6.5 FE results

The FE analysis was performed following the model and procedures described in the previous sections. Von Mises criterion were adopted in analysing results because the three dimensional constitutive model assumes that creep strain occurs when the elastic energy of distortion reaches a critical value. Figure 6.5 shows the deflection histories of the 8mm thick plate with 1, 2, and 3.3mm deflection in both the strain recovery (SR) and load recovery (LR) cases. For both cases, the stress relaxation process was carried out with fixed deflection of the plate at the target value for 8h. After the 8h stress relaxation process, for the SR case, the upper roller moved back to the origin point, so the deflection of the plate gradually recovered with time due to anelastic creep recovery. In the LR case, the upper roller moved up to the point where the force on the roller reached zero. The deflection of the plate was then maintained at this level to enable the generation of load due to the inhibition of anelastic creep recovery. Then the load was fully removed at 14h, and instant

springback occurred. The results show that the final deflections of the mid-point on the top surface were nearly the same in the SR and LR cases.

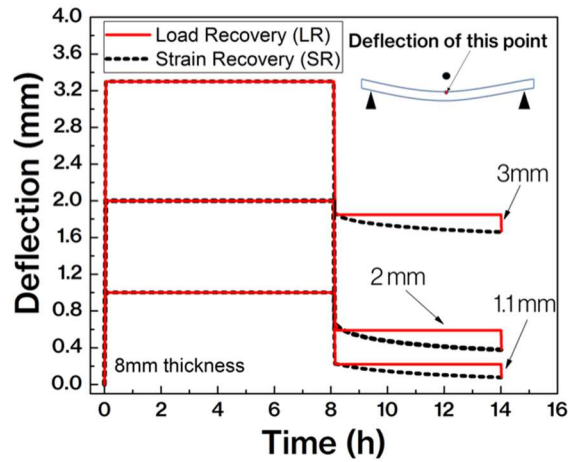


Figure 6.5. Deflection histories of both Strain recovery (SR, dashed lines) and Load recovery (LR, solid line) in FE simulation process.

Figure 6.6 shows the contour plots of the equivalent (von Mises) stress distributions of the stress relaxation stages of the SR and LR cases of 2mm deflection and 8mm thickness. Figure 6.6a shows the plate that was deformed elastically to 2mm deflection. At this stage, the maximum stress (232.7 MPa) can be observed at the centre of the top surface of the plate. After 4h of stress relaxation, as shown in Figure 6.6b, the overall stress distributed on the plate decreased and the maximum value fell to around 174 MPa. Figure 6.6c shows the stress distribution after 8h of stress relaxation. At this stage, the maximum stresses at the centres of the top and bottom surfaces dropped to about 145 MPa.

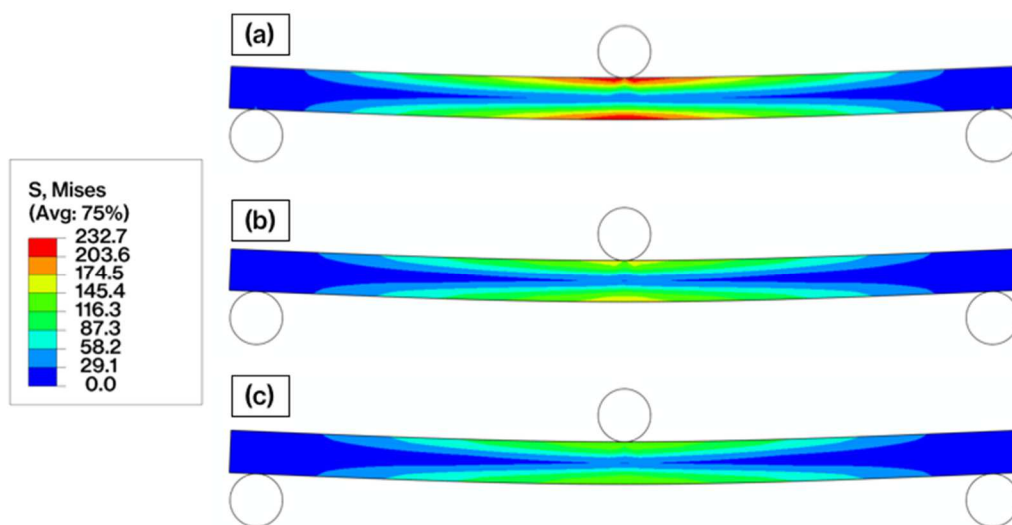


Figure 6.6. Equivalent (von Mises) stress distributions during stress relaxation of 8mm plate with 2mm deflection: (a) initial loading; (b) holding for 4h, (c) holding for 8h.

Equivalent (von Mises) stress distributions of the SR case of 2mm deflection and 8mm thickness in step 3 to step 4 are shown in Figure 6.7. As the applied load was removed from the plate, instant springback due to unloading occurred, and some of residual stress remained in the plate. Figure 6.7a shows the residual stress distribution after elastic unloading where the maximum residual stress was below 20 MPa.

In the strain recovery (SR) case, since the plate was in the ‘load-free’ condition, the residual stress on the plate gradually dissipated and anelastic creep recovery gradually proceeded. After 4h of strain recovery, the residual stress level slightly decreased to about 18 MPa, as shown in Figure 6.7b. Further holding did not reduce the stress significantly. At the end of strain recovery, the residual stress level was close to that of 4h, as shown in Figure 6.7c.

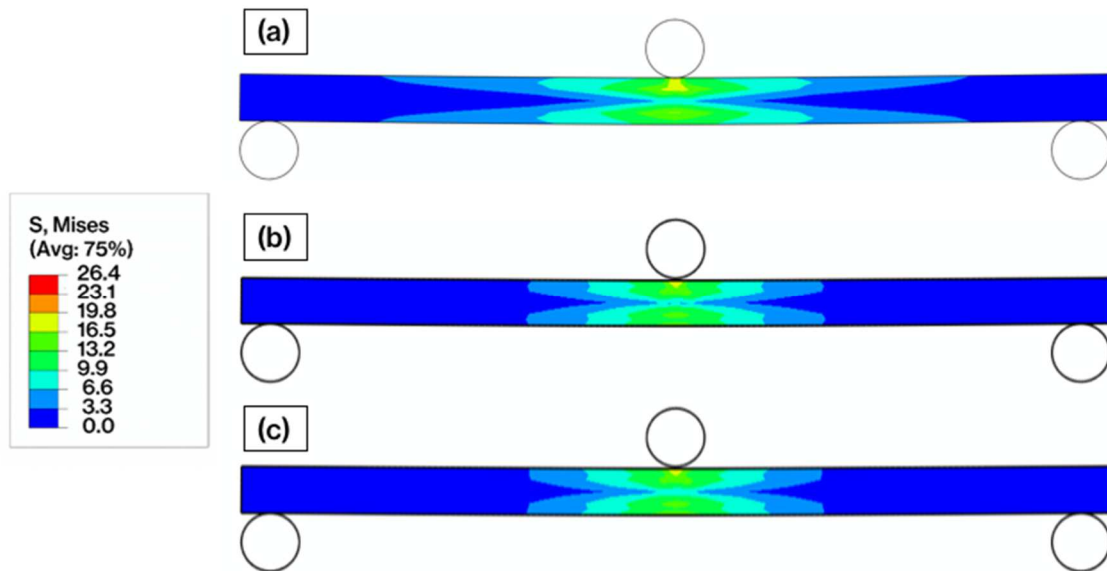


Figure 6.7. Equivalent (von Mises) stress distributions of the strain recovery (SR) case after stress relaxation: (a) Unloaded from 8h stress relaxation; (b) 4h of strain recovery, 12h total time; (c) 6h of strain recovery, 14h total time.

Figure 6.8 shows the equivalent (von Mises) stress distributions of the LR case in step 3 to step 5, since at the beginning of step 3 in the LR case, the conditions were the same as in the SR case, as shown in Figure 6.8a. After 4h of load recovery, internal stress gradually increased and the peak value reached 23 MPa, as shown in Figure 6.8b. At the end of the process, maximum internal stress reached about 26.4 MPa, as shown in Figure 6.8c. Figure 6.8d shows the residual stress distribution after full unloading in step 5.

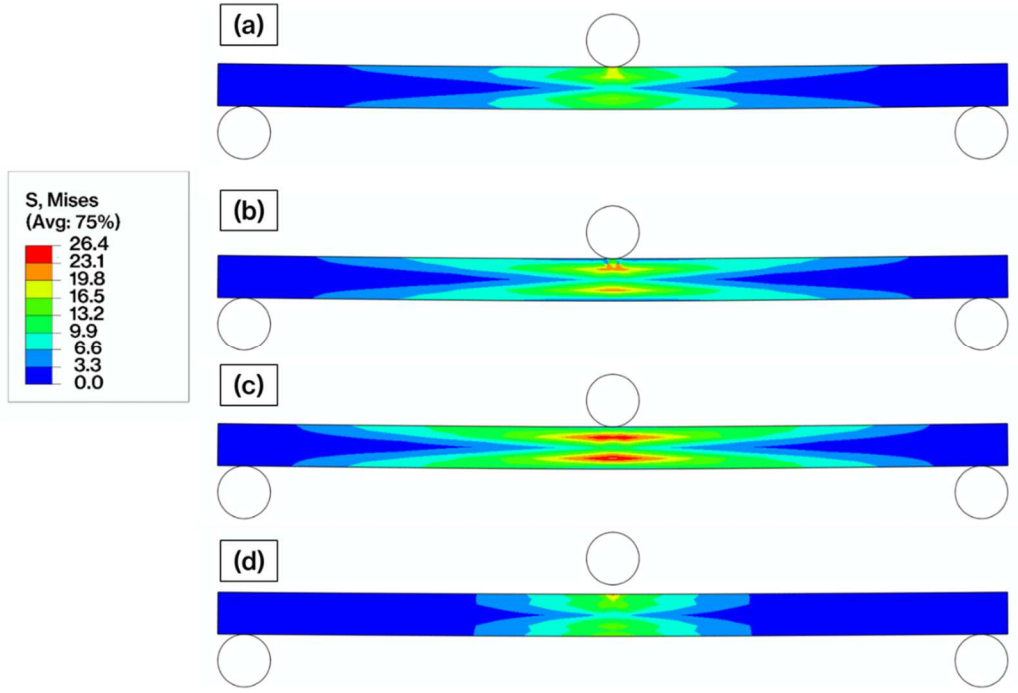


Figure 6.8. Equivalent (von Mises) stress distributions of the load recovery (LR) case after stress relaxation: (a) Unloaded from 8h stress relaxation; (b) 4h of load recovery, 12h total time; (c) 6h of load recovery, 14h total time; and (d) after full unloading in step 5.

Figure 6.9 shows the predicted distributions of the normalised mobile dislocation density (effective), normalised immobile dislocation density, normalised precipitate radius, yield strength, and back stress, after 8h of stress relaxation for both the SR and the LR cases. The contour plots of these parameters are very similar to those of the equivalent (Von Mises) stress distributions in Figure 6.6. This implies that the applied stress in creep has strong influence in affecting the growth of dislocations (mobile and immobile) and precipitates, which thereby influence the yield strength and back stress of the material. Most of the parameters, except yield strength, are largest in value at the centres of the top and bottom surfaces of the plate. The contour plots show that the value of the normalised mobile dislocation density (effective) in Figure 6.9a is significantly higher than the normalised immobile dislocation density in Figure 6.9b. The maximum values for the normalised mobile dislocation density (effective) and normalised immobile dislocation density are 0.14 and 0.051 respectively. The yield strength of the as-received material is 573 MPa, which drops to 434 MPa at the centres of the top and bottom surfaces of the plate, and 493 MPa for most of the plate. The maximum value of back stress is about 20 MPa at the top and bottom of the plate, as shown in Figure 6.9e. It should be noted that back stress (σ_{ij}^b) has the same sign as applied stress (σ_{ij}). In the constitutive equations, the relationship

between back stress (σ_{ij}^b) and applied stress (σ_{ij}), is $\sigma_{ij}^{eff} = \sigma_{ij} - n_0 \sigma_{ij}^b$. The negative sign in this equation changes the sign of the back stress and causes positive (tension) on the top surface of the beam.

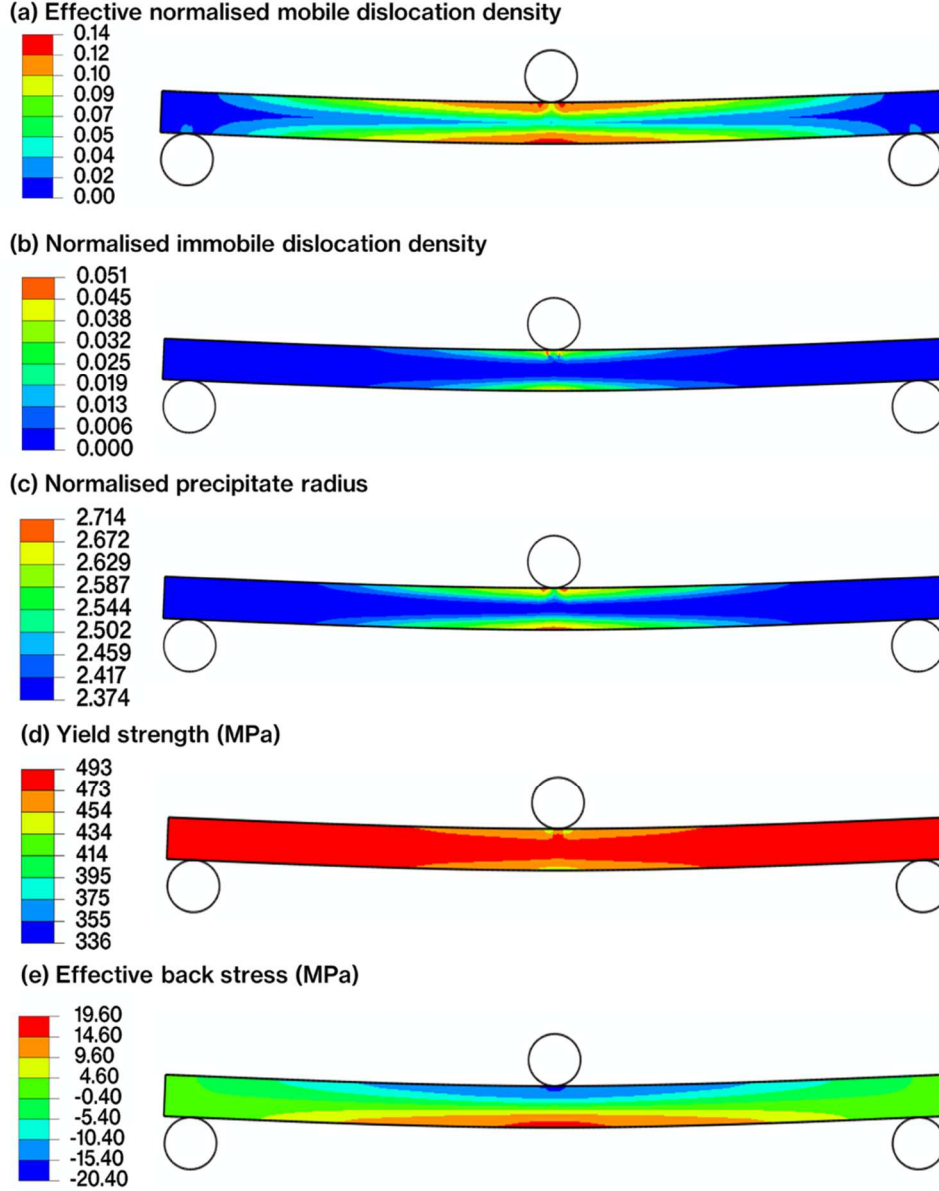


Figure 6.9. Predicted distributions of (a) Effective normalised mobile dislocation density, (b) normalised immobile dislocation density, (c) normalised precipitate radius, (d) yield strength, and (e) back stress after 8h of stress relaxation.

Figure 6.10 shows the contour plots of the same parameters in the LR case after 6h of load recovery (14h in total). In this stage, the normalised mobile dislocation density, Figure 6.10 a, and back stress, Figure 6.10e, dissipate significantly because they have transformed into load recovery force acting against the roller, as shown in Figure 6.11. The contour of the maximum normalised immobile dislocation density after load recovery, as shown in Figure

6.10b, changes little as compared with the contour after stress relaxation, as shown in Figure 6.9b, because in the load recovery stage, no extra external force is applied on the plate. The normalised precipitate size grows even further to a maximum of 3.352 at the centres of the top and bottom surfaces of the plate, as shown in Figure 6.9c. Yield strength evolution exhibits a softening behaviour, Figure 6.9d, because of the over aging effect; although the precipitate radius is increasing, the spacing between them also increases, so they become incoherent to the matrix. As a result, the contribution of the precipitates to the yield strength decreases.

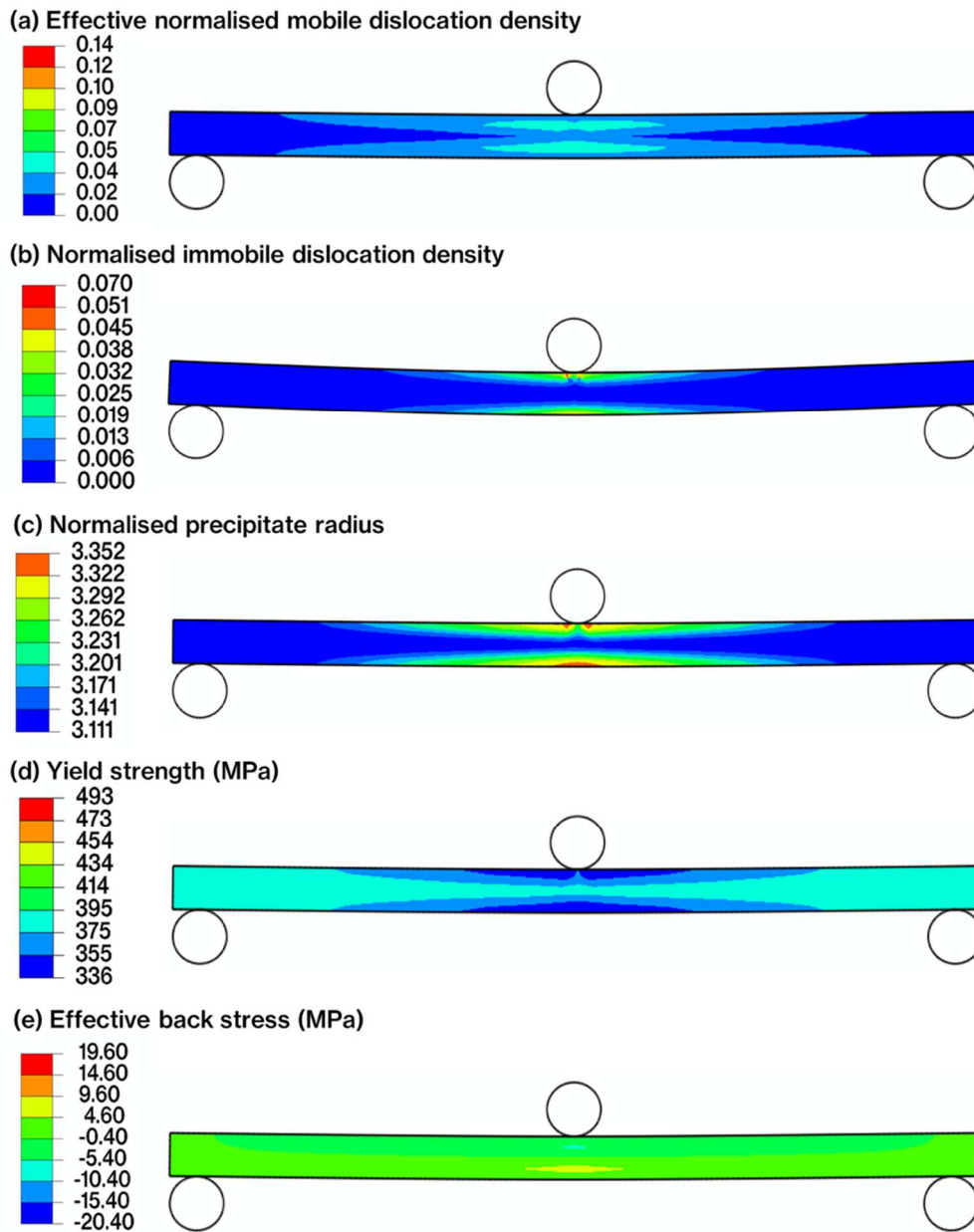


Figure 6.10. Predicted distributions of (a) normalised mobile dislocation density, (b) normalised immobile dislocation density, (c) normalised precipitate radius, (d) yield strength, and (e) back stress after 6h of load recovery (14h total time).

6.5.1 Comparison of experimental and FE results

Figure 6.11 shows the load variation histories of the upper roller throughout the stress relaxation and recovery process of the load recovery case. The experimental results presented in Chapter 3 were used to validate the FE predictions. It can be seen that close agreement can be achieved between experimental and FE prediction results for smaller deflections of the beam; e.g., 2mm deflection for 3mm thick plate, 3 and 4mm deflections for 5mm thick plate, and 1.1 and 2mm deflections for 8mm thick plate. The reason is that the higher the deflection, the more bending stress generated on the plate. Since the material constants for the constitutive equations were determined by uniaxial creep aging tests series, CART and CABT, for stress levels up to 162.5 MPa, the material constants might not be sufficient to predict cases with higher bending stresses on the plate.

A comparison of the springback measured from experiments and the predicted results from FE simulation is presented in Figure 6.12. It shows that for the plates of 3mm thickness with 2mm deflection, the predicted result is 3% higher than the experimental one for instant springback, and 1% after load recovery. When the deflection is 5.47 and 7.88mm, the predicted results are 6% and 9% lower than the experimental ones for the instant springback respectively, and 7% and 12% lower respectively for the springback after recovery, as shown in Figure 6.12a, b.

For plate of 5mm thickness and 2mm deflection, a 9% difference in instant springback can be observed for the prediction and experimental results. After recovery, the value increases to 19%. When the deflection increases to 3mm, the difference between prediction and experimental results becomes 1% for the instant springback and 4% after load recovery. As the deflection increases further to 4 and 4.9mm, the differences are 4% and 7% for the instant springback respectively, and 1% and 7% after load recovery, as shown in Figure 6.12c, d.

For the 7.88mm thick plate, when the deflection is below 2mm, the differences between prediction and experimental results are below 4% for instant springback and 8% after load recovery. When the deflection is 3.3mm, the predicted result becomes 20% lower than the experimental one, as shown in Figure 6.12e, f.

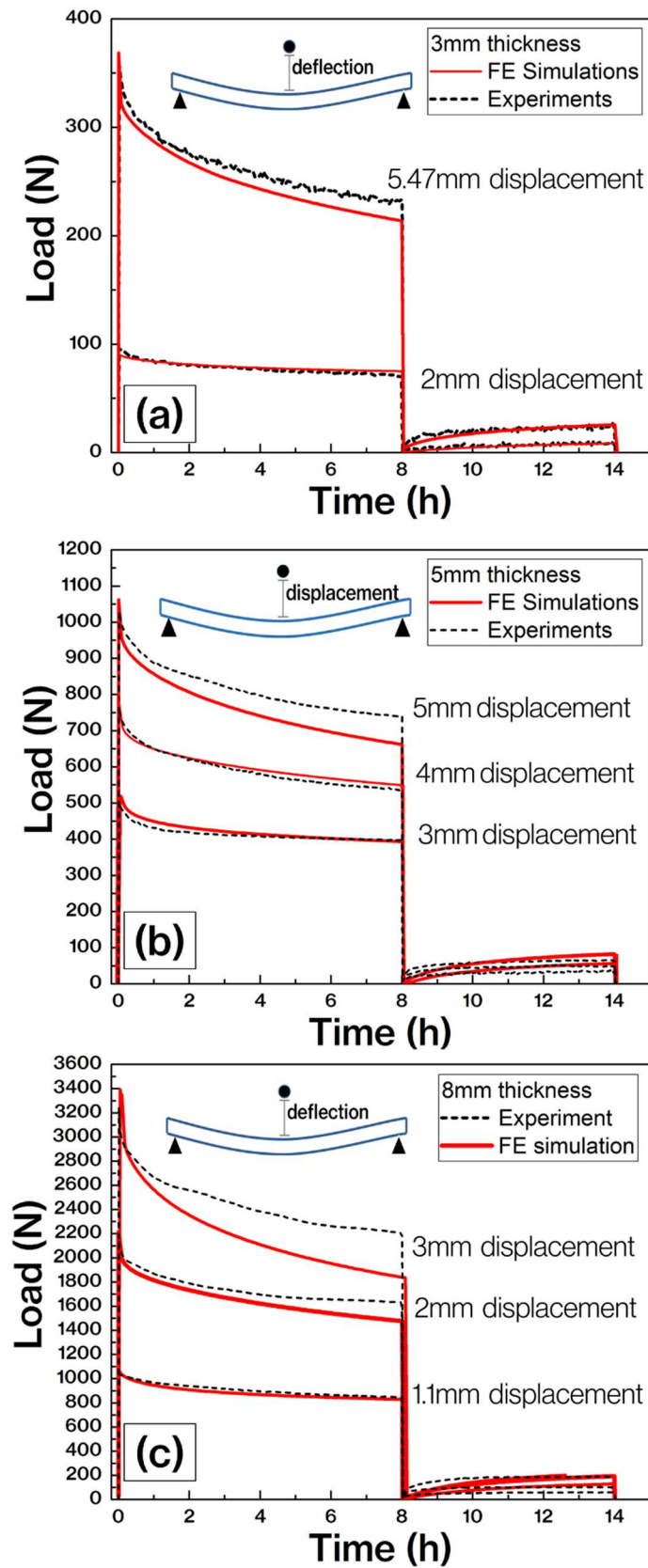


Figure 6.11. Experimental (dashed lines) and computed (solid lines) curves of three-point bending stress relaxation and back stress test for (a) 3mm, (b) 5mm, and (c) 8mm plate.

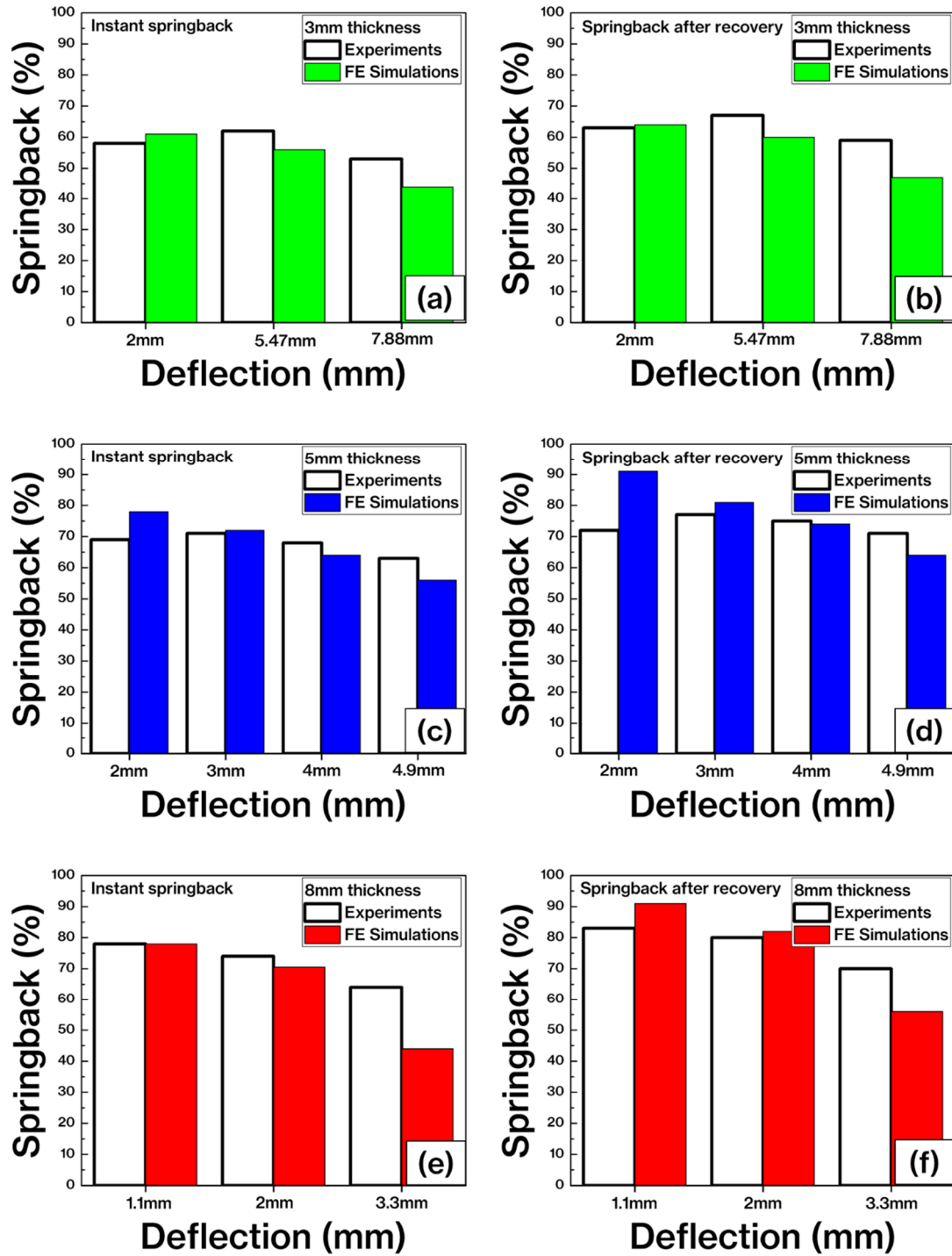


Figure 6.12. Comparison of springback from experimental measurement and FE prediction showing instant springback of (a) 3mm, (c) 5mm, and (e) 8mm thick plate, and springback after load recovery of (b) 3mm, (d) 5mm, and (f) 8mm thick plate.

6.6 Summary

The unified constitutive equations presented in the previous chapter have been integrated into an FE solver, ABAQUS/Explicit, to allow the simulation of not only the stress relaxation but also the anelastic creep recovery of deformation and load. Microstructural evolutions and other internal variables, namely, normalised mobile dislocation density, normalised immobile dislocation density, normalised precipitate radius, yield strength, and back stress, are also modelled.

The experimental results of three-point stress relaxation and recovery tests presented in chapter 4 have been compared with the FE predicted results. Good agreement in springback prediction can be achieved when the deflection of the plate is smaller than 5.47mm for 3mm plate, 4mm deflection for 5mm plate, and 2mm deflection for 8mm plate. When the deflection is higher than the values mentioned above, the higher bending stresses on the plate decrease the predictability, since this model was determined by uniaxial CART and CABT for stress levels of only up to 162.5 MPa.

7 CONCLUDING REMARKS

7.1 General conclusions

The work presented in this thesis was intended to characterise the properties of the AA7050-TAF alloy regarding deformation, strength, and microstructural evolution, and anelastic creep recovery behaviour in the creep age forming process, so as to allow constitutive modelling for the prediction of shape and springback of parts formed by creep age forming. Analysis of conventional creep tests and tensile tests revealed that the material possesses typical overaging behaviour, as characterised by the time and applied stress dependence. The applied stress in creep aging can decrease the total strength of the material. The difference in yield strength of the samples under different applied stresses was minimised after 8h of aging.

Part of the experimental work was concerned with the microstructures of the AA7050-TAF alloy, with particular attention to the evolution of the precipitates, which may be influenced by the stress during the creep age forming process. In this regard, microstructural analysis obtained from TEM demonstrated that the growth rate of GP zones and η' precipitates can be accelerated by applying stress, which also leads to faster coarsening of the η precipitates. It is also shown that the nucleation of GP zones is a continuous process that occurs even after 8h of pure aging and creep aging. The influence of evolution of the dislocation density on hardening, however, could not be detected with any certainty.

Great emphasis was placed on the role of anelasticity on the total deformation and springback of parts formed by the creep age forming process. The new experimental methods – the creep aging and recovery test (CART) and the creep aging and back stress test (CABT) – were designed to elucidate the problem by monitoring the anelastic strain and back stress after unloading from uniaxial creep aging. The three-point bending stress relaxation and back stress test (SRBT) was performed to simulate the creep age forming process. The results showed that if anelastic creep recovery is considered, springback can be 5-7% higher in general. Furthermore, the combination of observations from the literature and the experimental results suggests that the cause of anelastic creep recovery may be subgrain cell structures formed during the creep aging process.

Also of great interest is the ability to predict the shape, springback, and microstructural evolution of the parts formed in the process. On the basis of the micro-mechanisms mentioned above, a model has been created to introduce anelastic creep recovery into the scheme. The origins and capabilities of the constitutive model have been described in detail. Based on the results presented in Chapter 5, it is now possible to estimate the anelastic strain and back stress from the creep aging histories of the material. This in turn gives an indication of the normalised precipitate radius, mobile and immobile dislocation densities, solid solution strengthening, dislocation hardening, and precipitation hardening variables – the state variables that influence the mechanical properties of the material during the creep aging process.

A full multiaxial version of the constitutive model has been implemented in ABAQUS, a commercial FE software program, through a user-defined subroutine, VUMAT. An integrated numerical procedure to simulate the creep age forming process has been developed. The simulation is mainly concerned with the prediction of stress relaxation behaviour and the effect of anelasticity on springback of the formed parts in the creep age forming process. The predicted results from FE were compared with experimental results. The model proved its suitability by reproducing the stress relaxation and load recovery trends of experimental data when the deflection of the plate is smaller than 5.47mm deflection for 3mm plates, 4mm deflection for 5mm plates, and 2mm deflection for 8mm plates.

To return to the main aim of this study, the present work has covered the full range spanning basic mechanical and metallographic analysis to the development of constitutive models and FE simulations for industrial applications. The main findings and contributions are listed as follows:

- Through mechanical testing, it was discovered that the applied stress in creep aging can decrease the total strength of the aluminium alloy AA7050 TAF – the higher the applied stress, the lower the yield strength. The difference in yield strength under different applied stresses, however, is minimised after 8h of aging.
- Microstructural characterisation by TEM indicated that the influence of the material's yield strength from different applied stresses could be due to the faster coarsening rate of the η precipitates under higher applied stress in creep, accompanied by the acceleration of the growth rate of GP zones and η' precipitates.
- The concept of anelasticity has been introduced into the field of creep age forming because in this field, the main concern is the final shape after the forming process. Previous studies, however, generally used conventional creep tests, which do not consider the final deformation after unloading. The creep aging and recovery test (CART) was therefore performed to reveal how anelasticity can affect the final deformation.
- Most studies have pointed out that internal stresses are the cause of anelastic creep recovery. The creep aging and back stress test (CABT) has been developed in an attempt to measure the internal stress, or back stress, of a material after creep aging.
- The idea of monitoring the stress evolution on the formed parts during the creep age forming process was raised. An *in-situ* three-point bending stress relaxation and back stress test (SRBT) has been designed to evaluate the stress relaxation and load recovery behaviour of the formed parts for creep age forming.
- A new constitutive model considering anelastic creep recovery has been developed and implemented in ABAQUS, a commercial FE software program. The efficacy of this model has been validated by comparing the simulation results with the experimental results.

7.2 Future work

The major motivation for this work was the importance of achieving better understanding of the mechanical properties, micro-mechanisms, and prediction of the deformation of materials in the creep age forming process. For various reasons, however, some issues in this study still remain unsolved and need to be clarified. The following are suggestions for future research work:

- Through TEM analysis, the precipitates that form under pure aging and creep aging have been examined. How the applied stress in creep aging can influence the growth kinetics of precipitates needs to be characterised.
- A new testing technique, the creep aging and back stress (CABT) test, has been developed to measure the back stress of a material after unloading from creep. Further investigations are required to validate this technique to see if it can provide comparable results with conventional testing methods, i.e., incremental unloading stress relaxation tests and stress transient dip tests, for measuring internal stress.
- In this study, under an isothermal condition, the way anelasticity can influence the springback of the formed parts in the creep age forming process has been examined. In industrial practice, however, the formed parts are generally unloaded and cooled to room temperature after the forming process. The behaviour of anelastic creep recovery at room temperature needs to be further investigated.
- Experimental works and FE simulations on actual creep age forming process may also be performed to evaluate the effect of anelastic creep recovery on the springback of the formed parts after creep age forming, and to validate the constitutive equations developed in this study.
- There is a serious lack of available microstructural data, such as the factors in solid solution hardening, dislocation densities, and precipitate radius that are encountered in current constitutive models. This lack greatly limits the applicability of microstructural evolutions to the meaningful predictions.
- The constitutive model developed in this work is based on creep mechanisms. In the industrial creep age forming process, however, formed parts are sometimes loaded to above the yielding point. It is advisable to modify or extend the current model to cover

the physical mechanisms in plastic deformation and plastic flow rules to expand the scope of predictions.

7.3 Applications

The main scope for application of this study is to analyse the manufacturing problems involving the characterisation and prediction of microstructural evolutions, the strength loss due to overaging, and the time-dependent inelastic/anelastic material behaviours of materials. For this purpose, efforts have been made to develop new experimental techniques and constitutive modelling.

Experimental study revealed that anelastic strain may contribute up to 30% of the creep strain after 8h creep aging. This contribution implies that the prediction of total deformation determined by the conventional creep aging tests is larger than the actual creep age forming process. The CART and CABT developed in this study can provide complete data on total deformation or back stress from loading to unloading. They are more suitable than the conventional creep tests for investigating the total amount of permanent deformation after creep age forming. Therefore, they have the potential to become new standard experimental methods.

The SRBT developed in this study also presents a way to monitor the evolution of load applied on the formed parts during and after the creep age forming process with load cells. In industrial practice, it is advisable to record the load evolution during the forming process, since the load applied on the formed parts determines the degree of deformation.

The SRBTs have showed that anelastic creep recovery influences 5-7% of the springback under isothermal conditions. Although the ways that the cooling process and room temperature can affect anelasticity are still unclear, this study suggests that after the creep age forming process, the forms of anelastic creep recovery are either anelastic strain, which further recovers the original shapes of the formed parts, or back stress, which is stored in the formed parts. Those factors should be considered in the industrial creep age forming process.

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Appendix A. Calibration of Capacitance Gauges

The methodologies described in this appendix are the procedures for calibrating capacitance gauges and performing creep-aging tests.

- **Alignment of capacitance gauges**
 - The upper parts of capacitance gauges were adjusted to the same length, Figure A.1a. The specimen was fitted with the lower parts of capacitance gauges to see if the length was proper or not, Figure A.1b. A proper length means that there are small spaces between the lower and upper parts of the capacitance gauges.

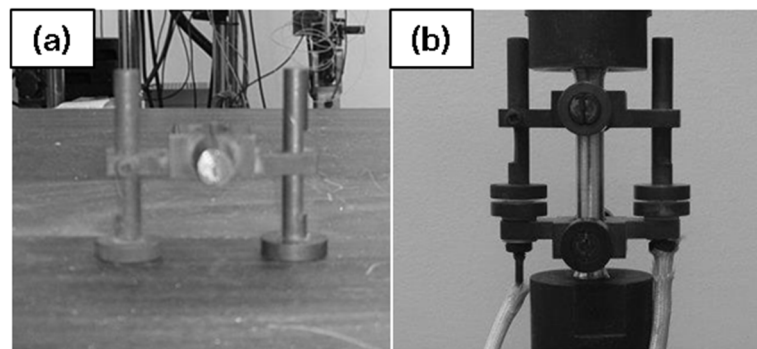


Figure A.1. (a) Upper parts of capacity gauges. (b) Aligned capacity gauges on the specimen.

- **Calibration of the capacitance sensors and test piece preparation**
 - The capacitance gauges were connected to the voltage converter and then mounted on an IHMEC calibrator, as shown in Figure A.2. The voltage measurements were taken on known displacement intervals. Rotating the wheel on the IHMEC calibrator changed the voltage reading with the displacement. A linear relationship was captured through this calibration.

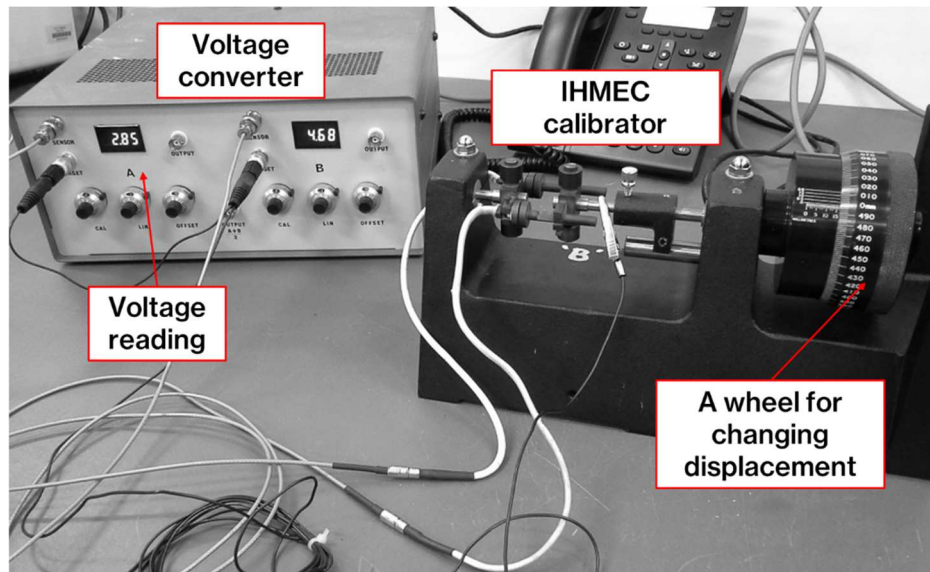


Figure A.2. Equipment set for calibration process.

- **Installation of thermocouples and capacitance gauges on the samples**

- The thermocouples were wired on the specimens between two ridges and the threads of the specimen. The rigs on the specimen were locked via the screw on the sample, as shown in Figure A.3.

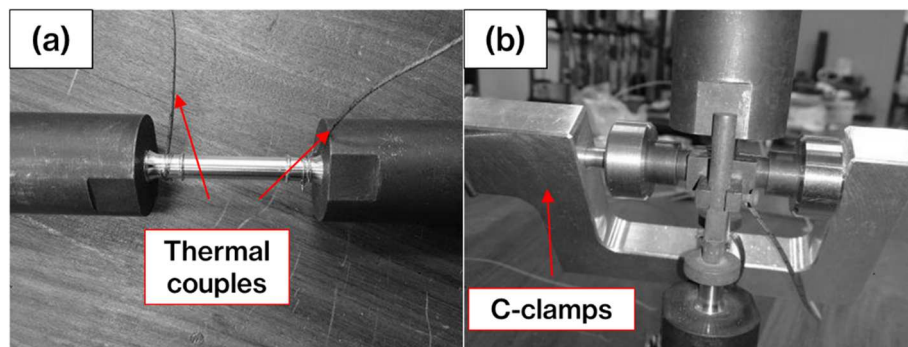


Figure A.3. (a) Sample with thermocouples. (b) Installation of gauges on the sample.

- **Setup the sample on the creep machine**

- The specimens with the rigs at the top and the end of the sample were fixed on the rods with pins. They were then placed into the furnace and locked onto the creep machine with T-pins. The configuration of the testing machine after setting up can be seen in Figure A.4 below.
- Two cables on the lower parts of capacitance gauges were then connected to the converter. The clips on the voltage converter were mounted on the lower T-pin

to complete a circuit. Then two pairs of cables on the voltage converter then linked to the computer for data collection and management.

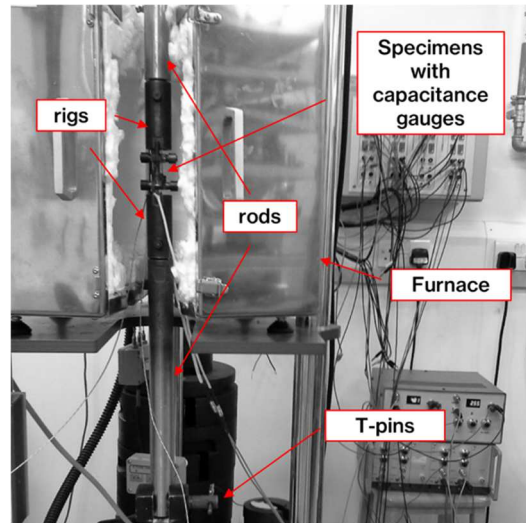


Figure A.4. Configuration of the testing machine for uniaxial creep aging tests after setting up.

- **Temperature and load configuration of testing machine**

- For the material in the as-received condition, the temperature needed for creep aging tests is 174°C. The temperature variations during heating process can be seen in Figure A.5, where the heating process to 174°C takes about 90 min.
- When the temperature reached 174°C, the loading process started and the furnace was set to maintain the temperature during the whole testing period. The loading method involved putting the balancing weight on the side of the lever arm of the creep machine opposite to the specimen.

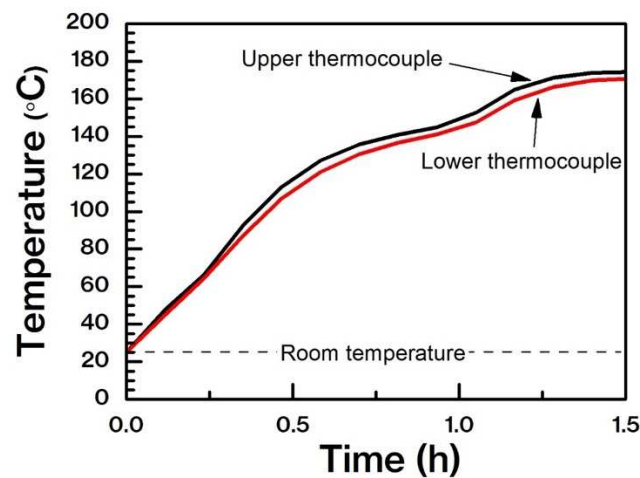


Figure A.5. Heating programme of the specimen during heating up process in pure aging and creep aging tests.

Appendix B. Subroutine of the constitutive model

This appendix presents the user defined subroutine VUMAT for the constitutive model described in Chapter 5. The code is written in Fortran 77.

```
subroutine vumat(  
  
* Read only (unmodifiable)variables -  
  
1  nblock, ndir, nshr, nstatev, nfieldv, nprops, lanneal,  
  
2  stepTime, totalTime, dt, cmname, coordMp, charLength,  
  
3  props, density, strainInc, relSpinInc,  
  
4  tempOld, stretchOld, defgradOld, fieldOld,  
  
5  stressOld, stateOld, enerInternOld, enerInelasOld,  
  
6  tempNew, stretchNew, defgradNew, fieldNew,  
  
* Write only (modifiable) variables -  
  
7  stressNew, stateNew, enerInternNew, enerInelasNew )  
  
*  
  
include 'vaba_param.inc'  
  
*  
  
dimension props(nprops), density(nblock), coordMp(nblock,*),  
  
1  charLength(nblock), strainInc(nblock,ndir+nshr),  
  
2  relSpinInc(nblock,nshr), tempOld(nblock),  
  
3  stretchOld(nblock,ndir+nshr),  
  
4  defgradOld(nblock,ndir+nshr+nshr),
```

```

5 fieldOld(nblock,nfieldv), stressOld(nblock,ndir+nshr),
6 stateOld(nblock,nstatev), enerInternOld(nblock),
7 enerInelasOld(nblock), tempNew(nblock),
8 stretchNew(nblock,ndir+nshr),
8 defgradNew(nblock,ndir+nshr+nshr),
9 fieldNew(nblock,nfieldv),
1 stressNew(nblock,ndir+nshr), stateNew(nblock,nstatev),
2 enerInternNew(nblock), enerInelasNew(nblock)

```

*

```

Real*8 dy,SMETAL,temp1,temp2,temp3,temp4,temp5,VMS,Tdis,
3 EStrain,deltaStress,dummy

integer g,r,ElastoFlag

dimension dy(nblock,45),SMETAL(21),Tdis(5),
3 EStrain(nshr+ndir),deltaStress(ndir+nshr)

character*80 cmname

```

SMETAL(1) =9.87e-06

SMETAL(2) =0.054

SMETAL(3) =6.6

SMETAL(4) =316

SMETAL(5) =2988

SMETAL(6) =6

SMETAL(7) =124.8

SMETAL(8) =1.07

SMETAL(9) =0.037

SMETAL(10) =5.0

SMETAL(11) =1.3

SMETAL(12) =2.4

SMETAL(13) =80.2

SMETAL(14) =0.97

SMETAL(15) =50

SMETAL(16) =0.001

SMETAL(17) =10

SMETAL(18) =0.1

SMETAL(19) =40.7

SMETAL(20) =56065.0

* SMETAL(20) =62000.0

* SMETAL(20) =60000.0

SMETAL(21) =0.34

C =====

c Determine If Run Elastic

ElastoFlag=0

IF (totalTime .le.0) ElastoFlag=1

IF (totalTime.le.0.1) ElastoFlag=1

IF (totalTime.ge.8.0999.and.totalTime.le.8.2001) ElastoFlag=2

c ElastoFlag=1

C *****

* —Loop for all elements —

DO 910 i = 1,nblock

* kill these later

StateNew(i,46)=dble(ndir)

StateNew(i,47)=dble(nshr)

C =====

C Code Goes Here

IF (ElastoFlag.eq.1) then

c_____

c Initialise

DO 101 r=1,ndir+nshr

 EStrain(r)=(strainInc(i,r))

101 continue

 EStar = dble(SMETAL(20)/((1.0+SMETAL(21)) *

 1 (1.0-(2.0*SMETAL(21)))))

C b. Diagonal components of D matrix

 vD = dble(1.0 - SMETAL(21))

C c. Indirect components of D matrix

 vI = dble(SMETAL(21))

C d. Shear components of D matrix

 vS = 2.0*dble((1.0 - (2.0*SMETAL(21)))/2.0)

 temp = vD*EStrain(1) + vI*EStrain(2) + vI*EStrain(3)

 Stressnew(i,1)=Stressold(i,1)+ EStar*temp

temp = vD*EStrain(2) + vI*EStrain(1) + vI*EStrain(3)

Stressnew(i,2)=Stressold(i,2)+ EStar*temp

temp = vD*EStrain(3) + vI*EStrain(1) + vI*EStrain(2)

Stressnew(i,3)=Stressold(i,3)+ EStar*temp

DO 301 r=1,nshr

Stressnew(i,r+3) = StressOld(i,r+3)+EStar*vS*EStrain(r+3)

301 continue

DO 102 g = 1,45

StateNew(i,g)=1e-6

102 continue

StateNew(i,30)=273.0

StateNew(i,28)=1.0

c_____

c Continue Elastic

ELSE IF (ElastoFlag.eq.2) THEN

DO 103 r=1,ndir+nshr

$$E\text{Strain}(r)=(\text{strainInc}(i,r))$$

103 continue

$$E\text{Star} = \text{dble}(\text{SMETAL}(20)/((1.0+\text{SMETAL}(21)) * \\ 1 (1.0-(2.0*\text{SMETAL}(21))))))$$

C b. Diagonal components of D matrix

$$vD = \text{dble}(1.0 - \text{SMETAL}(21))$$

C c. Indirect components of D matrix

$$vI = \text{dble}(\text{SMETAL}(21))$$

C d. Shear components of D matrix

$$vS = 2.0*\text{dble}((1.0 - (2.0*\text{SMETAL}(21)))/2.0)$$

$$\text{temp} = vD*E\text{Strain}(1) + vI*E\text{Strain}(2) + vI*E\text{Strain}(3)$$

$$\text{Stressnew}(i,1)=\text{Stressold}(i,1)+ E\text{Star}*\text{temp}$$

$$\text{temp} = vD*E\text{Strain}(2) + vI*E\text{Strain}(1) + vI*E\text{Strain}(3)$$

$$\text{Stressnew}(i,2)=\text{Stressold}(i,2)+ E\text{Star}*\text{temp}$$

$$\text{temp} = vD*E\text{Strain}(3) + vI*E\text{Strain}(1) + vI*E\text{Strain}(2)$$

$$\text{Stressnew}(i,3)=\text{Stressold}(i,3)+ E\text{Star}*\text{temp}$$

DO 302 r=1,nshr

Stressnew(i,r+3) = StressOld(i,r+3)+EStar*vS*EStrain(r+3)

302 continue

DO 104 g = 1,45

StateNew(i,g)=StateOld(i,g)

104 continue

c=====

c Creep

ELSE

c-----

c1 Update Effective Stress – Eq(2)

DO 105 r=1,ndir+nshr

StateNew(i,6+r)=stressOld(i,r)-SMETAL(3)*Stateold(i,18+r)

105 continue

c-----

c2 Update Von Mises Effective Stress – Eq(4)

DO 106 r=1,ndir+nshr

StateNew(i,32+r) = StateNew(i,6+r)

106 continue

dummy=(1.0/3.0)*(StateNew(i,6+1)+StateNew(i,6+2)+StateNew(i,6+3))

DO 107 r=1,ndir

StateNew(i,32+r) = StateNew(i,32+r) - dummy

107 continue

c_____

c3 Update Von Mises Effective Stress – Eq(5)

C a. direct stress components

temp1 = (StateNew(i,7)-StateNew(i,8))**2.0

temp2 = (StateNew(i,8)-StateNew(i,9))**2.0

temp3 = (StateNew(i,9)-StateNew(i,7))**2.0

temp4 = 0.5*(temp1+temp2+temp3)

C b. add shear stress components

temp4 = temp4 + 3.0*(StateNew(i,10)**2.0+StateNew(i,11)**2.0

2 +StateNew(i,12)**2.0)

C c. vmises equivalent stress

StateNew(i,26) = sqrt(temp4)

c_____

c4 Creep Change Eq(1)

dy(i,25) = SMETAL(1)*SINH(SMETAL(2)*(StateNew(i,26)

2*(1.-Stateold(i,27))))

StateNew(i,25)=Stateold(i,25)+dy(i,25)*dt

c_____

c5 Creep Change Eq(3)

if (StateOld(i,26).le.(1e-6)) then

DO 108 r=1,ndir+nshr

dy(i,r) =0.0

108 continue

else

DO 109 r=1,ndir+nshr

```

dy(i,r) = dy(i,25)*(3.0/2.0)*StateNew(i,6+r)/StateOld(i,26)

109 continue

end if

c   +++

DO 110 r=1,ndir+nshr

    StateNew(i,r)=Stateold(i,r)+dy(i,r)*dt

110 continue

c_____

c6  Eq(6)

DO 111 r=1,ndir+nshr

    if (abs(Stateold(i,12+r)).le.1e-12) stateold(i,12+r)=1e-12

    dy(i,12+r) =

2   SMETAL(4)*(1-abs(stateold(i,12+r)))*dy(i,r)-

2   sign(1.0,stateold(i,12+r))*(SMETAL(5)*

2   (abs(stateold(i,12+r))**SMETAL(6)))

```

StateNew(i,12+r)=Stateold(i,12+r)+dy(i,12+r)*dt

c if (StateNew(i,12+r).le.1e-8) then

c StateNew(i,12+r)=1e-6

c dy(i,12+r)=0.0

c end if

111 continue

c_____

c7 Update effective dislocations (Von Mises) Eq(9)

DO 112 r=1,ndir+nshr

StateNew(i,39+r)= StateNew(i,12+r)

112 continue

dummy=(1.0/3.0)*(StateNew(i,13)

2 +StateNew(i,14)+StateNew(i,15))

DO 113 r=1,ndir

StateNew(i,39+r)= StateNew(i,39+r)-dummy

113 continue

c—————

c8 Eq(10)

C Calculate equivalent components

C a. direct M dislocation components

Tdis(1) = (StateNew(i,13)-StateNew(i,14))**2.0

Tdis(2) = (StateNew(i,14)-StateNew(i,15))**2.0

Tdis(3) = (StateNew(i,15)-StateNew(i,13))**2.0

Tdis(4) = 0.5*(Tdis(1)+Tdis(2)+Tdis(3))

C b. shear components

Tdis(5) = 0.0000

Tdis(5) = Tdis(5) + 3.0*(StateNew(i,16)**2.0+StateNew(i,17)**2.0

2 +StateNew(i,18)**2.0)

C c. Equivalent dislocation

$$\text{StateNew}(i,27) = \text{sqrt}(\text{Tdis}(4)+\text{Tdis}(5))$$

c_____

c9 Eq(7)

$$\text{if } (\text{abs}(\text{Stateold}(i,12+r)) \leq 1e-12) \text{ stateold}(i,12+r)=1e-12$$

DO 114 r=1,ndir+nshr

$$\text{dy}(i,18+r) =$$

$$2(\text{SMETAL}(7)*\text{SMETAL}(8)*\text{abs}(\text{StateOld}(i,12+r))^{**}(\text{SMETAL}(8)-1.0)$$

$$2 * \text{dy}(i,12+r))$$

$$\text{StateNew}(i,18+r)=\text{Stateold}(i,18+r)+\text{dy}(i,18+r)*\text{dt}$$

114 continue

c_____

c10 Eq(8)

$$\text{dy}(i,28) = \text{SMETAL}(9)*((\text{SMETAL}(10)-$$

$$2 \text{ StateOld}(i,28))^{**}\text{SMETAL}(11))*(1+\text{SMETAL}(19)*((\text{StateOld}(i,27))$$

$$3 **\text{SMETAL}(12)))$$

$$\text{StateNew}(i,28)=\text{Stateold}(i,28)+\text{dy}(i,28)*\text{dt}$$

$$\text{if } (\text{StateNew}(i,28) \leq 1e-8) \text{ then}$$

StateNew(i,28)=1e-6

end if

c_____

c11 Eq(11)

$dy(i,30) = SMETAL(13)*dy(i,28)**SMETAL(14)*(1-StateOld(i,28))$

$StateNew(i,30)=Stateold(i,30)+dy(i,30)*dt$

c_____

c12 Eq(12)

$dy(i,29) = SMETAL(15)*StateNew(i,27)*Stateold(i,5)-SMETAL(16)$

$2*StateOld(i,29)**(1.5)$

$StateNew(i,29)=Stateold(i,29)+dy(i,29)*dt$

c_____

c13 Eq(13)

$dy(i,31) = SMETAL(17)*SMETAL(18)*StateOld(i,29)$

$2**(SMETAL(18)-1)*dy(i,29)$

$StateNew(i,31)=Stateold(i,31)+dy(i,31)*dt$

c_____

c14 Eq(14)

$$\text{StateNew}(i,32) = 300.0 + \text{SQRT}(\text{StateNew}(i,30)**2 \\ + \text{StateNew}(i,31)**2)$$

c_____

cLast Calculate the Stress Components

DO 115 r=1,ndir+nshr

$$\text{EStrain}(r) = (\text{strainInc}(i,r) - dy(i,r)*dt)$$

115 continue

$$\text{EStar} = \text{dble}(\text{SMETAL}(20) / ((1.0 + \text{SMETAL}(21)) * \\ 1 (1.0 - (2.0 * \text{SMETAL}(21)))))$$

C b. Diagonal components of D matrix

$$vD = \text{dble}(1.0 - \text{SMETAL}(21))$$

C c. Indirect components of D matrix

$$vI = \text{dble}(\text{SMETAL}(21))$$

C d. Shear components of D matrix

$$vS = 2.0 * \text{dble}((1.0 - (2.0 * \text{SMETAL}(21))) / 2.0)$$

$$\text{temp} = vD * \text{EStrain}(1) + vI * \text{EStrain}(2) + vI * \text{EStrain}(3)$$


```
Stressnew(i,1)=Stressold(i,1)+ EStar*temp
```

```
temp = vD*EStrain(2) + vI*EStrain(1) + vI*EStrain(3)
```

```
Stressnew(i,2)=Stressold(i,2)+ EStar*temp
```

```
temp = vD*EStrain(3) + vI*EStrain(1) + vI*EStrain(2)
```

```
Stressnew(i,3)=Stressold(i,3)+ EStar*temp
```

```
DO 303 r=1,nshr
```

```
    Stressnew(i,r+3) = StressOld(i,r+3)+EStar*vS*EStrain(r+3)
```

```
303 continue
```

```
C=====
```

```
ENDIF
```

```
C=====
```

```
910 continue
```

```
return
```

end

C *****

c Subroutine To Calculate Stresses

SUBROUTINE CalculateStresses(EStrain,Stressnew,Stressold,

1 SMETAL,ndir,nshr,nblock,i)

integer ndir,nshr,nblock,i,r

Real*8 SMETAL,EStrain,Stressnew,EStar,vD,vI,vS,temp

dimension SMETAL(21), EStrain(nshr+ndir),

1 Stressnew(nblock,ndir+nshr),Stressold(nblock,ndir+nshr)

EStar = dble(SMETAL(20))/((1.0+SMETAL(21)) *

1 (1.0-(2.0*SMETAL(21))))

C b. Diagonal components of D matrix

vD = dble(1.0 - SMETAL(21))

C c. Indirect components of D matrix

vI = dble(SMETAL(21))

C d. Shear components of D matrix

vS = 2.0*dble((1.0 - (2.0*SMETAL(21)))/2.0)

temp = vD*EStrain(1) + vI*EStrain(2) + vI*EStrain(3)

Stressnew(i,1)=Stressold(i,1)+ EStar*temp

temp = vD*EStrain(2) + vI*EStrain(1) + vI*EStrain(3)

Stressnew(i,2)=Stressold(i,2)+ EStar*temp

temp = vD*EStrain(3) + vI*EStrain(1) + vI*EStrain(2)

Stressnew(i,3)=Stressold(i,3)+ EStar*temp

DO 201 r=1,nshr

Stressnew(i,r+3)= StressOld(i,r+3)+EStar*vS*EStrain(r+3)

201 continue

RETURN

END