

THE ORIGIN OF DEFECTS DURING AL-MG-SI EXTRUSION

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Abstract

The material flow and surface generation of an Al-Mg-Si alloy (AA6063) have been studied. The structure and sub-structure at each stage of the extrusion process were characterised, both for the direct and indirect mode of extrusion, utilising transmission and scanning electron microscopy augmented with optical microscopy. The effect of the process parameters on the resulting extrudate structure were quantified.

The development of structure within the deformation zones of partially extruded billets, under varying process conditions, was established and the pressure requirements for rod and shaped extrusions were investigated.

The hot torsion test was developed and used to formulate constitutive equations, two of which possessed strain dependence. The Zener-Hollomon equation was subsequently applied to the extrusion process. The structures produced in this test were compared with those found in extrusion at congruous temperature and strain-rate, or temperature compensated strain-rate conditions. The hot torsion test was subsequently validated as a constitutive and structural modelling technique for the extrusion process.

The mechanisms of generation of extrudate surface was established together with the origin of surface defects formed during the extrusion process with the aid of specially designed 'split-die' tooling. The most deleterious of surface defects formed in the architectural alloy AA6063, pick-up and die lines, was studied and a model for the mechanical and structural mechanisms of formation proposed. The surface structure was characterised using optical, transmission (with modified specimen preparation technique) and scanning (B.E.I.) electron microscopy and the topographical features using stylus and interferometric measurement techniques coupled with scanning (S.E.I.) electron microscopy. The effect of die land length on the mechanical formation mechanism of pick-up and die lines was investigated with the aid of a series of dies with varying die land lengths, including knife edge.

The effect of process conditions on the tensile properties of AA6063 in the -O, -T1, -T4 and -T6 conditions was evaluated. A limit diagram for AA6063 is presented with structural information. Also presented are the modifications to the limit diagram for the representation of topological features.

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THE ORIGIN OF DEFECTS DURING AL-MG-SI EXTRUSION

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CHAPTER ONE

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Introduction

At the turn of the Nineteenth Century, Bramah invented an hydraulic extrusion press for the production of lead pipes. By the end of the Nineteenth Century, improvements in press design and hot working steel enabled Alexander Dick to successfully apply the process to higher melting point materials. Today, extrusion is one of the principal metal forming processes. The main advantage of extrusion is that complex sections, both solid and hollow, can be manufactured in a single operation, which meet strict geometric, property and cosmetic goals. The relative importance of these goals varies, depending on the product application.

The most commonly extruded aluminium alloys belong to the AA6XXX series. These alloys contain magnesium and silicon which combine to form the age hardening constituent Mg_2Si . A large portion of the extrusion products' market is for architectural and decorative trim application. The alloy which satisfies the majority of this market is AA6063.

An added advantage of the extrusion process, in extruding AA6063, is that with suitable homogenisation practices the magnesium and silicon can be retained in solution by quenching the extrudate as it emerges from the die. This makes the need for a separate solution heat treatment redundant, thereby improving the process economics.

considerably. (Retention of the solute is possible in AA6063 with forced air quenching.) The extrudate is thus in a suitable condition for natural or artificial ageing.

The work presented in this thesis examines the hot working behaviour of AA6063 during direct and indirect extrusion and determines the effects of process conditions on the structural and sub-structural features. The hot torsion test was utilised as a model deformation process, to establish constitutive equations for use in the extrusion analysis, using hot working theory.

Probably the most critical feature of AA6063 alloy extrusion is the attainment of an excellent surface finish, free from all surface defects. Despite the serious commercial implication of good quality surface generation, there is surprisingly little published research in the technical literature. In the present research, the formation mechanisms of the as-extruded surface defects are examined with special attention given to 'pick-up and die lines'.

CHAPTER TWO

CHAPTER TWO

Literature Review

2.1. Aluminium-Magnesium-Silicon

The use of magnesium (Mg) and silicon (Si) as alloying elements can be traced to 1918 with several Swiss experiments. Eight years later in 1926, the commercial beginning of Al-Mg-Si alloys was marked by the introduction of 'Aldur'. Today, the Al-Mg-Si system, designated the AA6XXX series by the Aluminium Association, is the most widely used alloy in the extruded products market. The AA6XXX series alloys have the following qualities; good corrosion resistance, surface finish, formability and medium strength; thus making them suitable for decorative architectural sections and structural applications.

The formation of the stoichiometric compound, magnesium silicide (Mg_2Si), makes the AA6XXX series alloys heat treatable and capable of achieving medium strength in the T6 condition.

There are two main alloys of the AA6XXX series used to meet consumer demands. These are AA6061, for medium strength, structural applications, and AA6063 for architectural applications. The other major alloys of the series include AA6351 and AA6005. (The chemical composition limits and tensile properties of the above alloys are given in Tables (2.1.) and (2.2.))(19).

Of the alloys mentioned above, AA6063 is considered to be the 'bread and butter' ⁽²⁰⁾ of the extrusion industry, possessing excellent extrusion characteristics and finishing capabilities. The Aluminium Association's chemical composition limits of AA6063 have a broad range for alloy and impurity elements from which a spectrum of strength and finishing properties can be developed. Three main divisions are offered by aluminium founders ⁽²¹⁾: these have magnesium silicide contents of 1, 0.8, and 0.7, for high strength, general purpose and high extrudability, respectively. The effects of magnesium silicide content and other alloying elements are discussed in the following sections.

2.2. The Effect of Alloying Elements

2.2.1. Magnesium Silicide (Mg_2Si)

Magnesium and silicon are the main alloying elements in the AA6XXX series and combine to form the stoichiometric constituent magnesium silicide (Mg_2Si), which is the primary hardening phase. The proportion of magnesium to silicon required for equilibrium magnesium silicide is 1.73:1. In a balanced alloy, aluminium and magnesium silicide can be treated as a quasi-binary system, thereby simplifying the phase diagram representation (Figure (2.1.)). The diagram shows that the maximum solubility of Mg_2Si in aluminium is 1.85wt% at 600°C.

An increase in the Mg_2Si content results in improved tensile

Table 2.1.

5.

Alloy	6005	6061	6063	6351
Si	0.6-0.9	0.4-0.8	0.2-0.6	0.7-1.3
Fe	0.35	0.70	0.35	0.50
Cu	0.10	0.15-0.40	0.10	0.10
Mn	0.10	0.15	0.10	0.4-0.8
Mg	0.4-0.6	0.8-1.2	0.45-0.90	0.4-0.8
Cr	0.10	0.04-0.35	0.10	-
Zn	0.10	0.25	0.10	0.20
Ti	0.10	0.15	0.10	0.20
Other	0.15	0.15	0.15	0.15

Table 2.2.

Alloy and Temper	UTS Nm ⁻²	Yeild Nm ⁻²
6005-T1	172	103
-T5	262	241
6061-0	152	110
-T1	179	97
-T4	179	110
-T5	-	-
-T6	262	241
6063-0	131	-
-T1	117	62
-T4	131	69
-T5	152	110
-T6	206	172
6351-T54	206	138

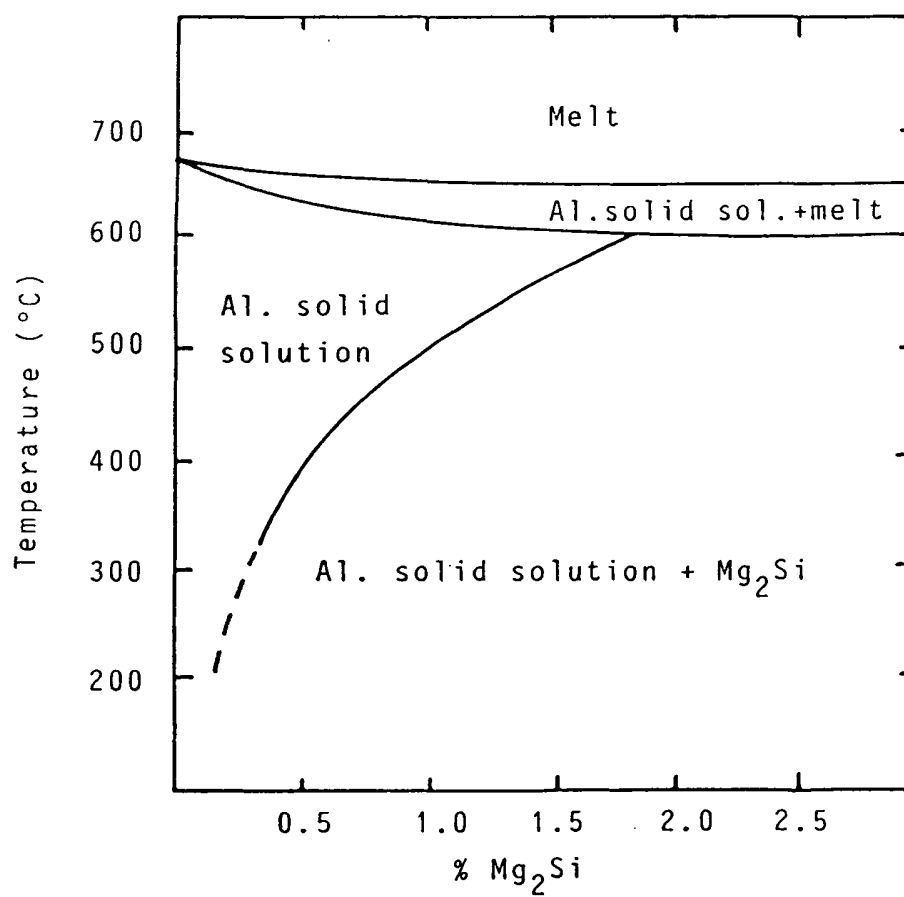


Figure 2.1 Phase Diagram of Al-Mg₂Si

properties⁽²²⁻²⁷⁾, but is accompanied by a loss of extrudability. For example, Traenkner⁽²⁸⁾ found that by increasing the Mg_2Si content from 0.5wt% to 0.95wt% and 1.35wt%, ram speed reductions of 40% and 70% respectively were necessary. However, when overall productivity is considered, increased Mg_2Si content can improve productivity by reducing the storage time required between extrusion and artificial ageing⁽³²⁾. For example, it has been reported⁽³³⁾ that for alloys with low content ($<0.35\% Mg_2Si$) the storage time may be several days, which is not feasible with the high production requirements of modern extrusion plants. The reason for the storage time is not fully understood but it is believed to be associated with the growth of very fine precipitates during this period⁽⁹⁹⁾.

To obtain full T6 age hardened properties, without the need of a separate solution treatment⁽²⁹⁾, the Mg_2Si must dissolve into solution during extrusion as the deforming material passes through the deformation zone. The Mg_2Si can then be retained in solution by quenching directly after extrusion. To ensure all the Mg_2Si dissolves during extrusion the precipitates must be fine and uniformly dispersed, as large precipitates require extended soak times for dissolution to complete. A fine and uniform distribution of Mg_2Si can be obtained by good casting processes⁽³⁰⁾, homogenisation (and subsequent cooling practices)^(23,29-31) and rapid reheating to the extrusion temperature⁽²⁰⁾. Minor alloying elements^(24,25,29), which form nucleation sites for Mg_2Si , also aids a fine distribution of precipitates.

Rapid precipitation of Mg_2Si occurs in the temperature range, $316^{\circ}C$ to $417^{\circ}C$ ⁽²⁷⁾ and will result in coarse precipitates unless this

temperature range is traversed rapidly on reheating to extrusion temperature. Coarse Mg_2Si particles cause low hardness⁽³⁰⁾, increased pick-up⁽³⁴⁾ on the extrude surface, and a cloudy finish upon anodising⁽³⁴⁻³⁶⁾.

For best strength and anodised brightness a fine and uniformly distributed Mg_2Si precipitate size of approximately $0.25\mu m$ is recommended⁽³⁶⁾.

An Mg_2Si precipitate size of $0.5\mu m$ is reported^(35,37) to give the best compromise of extrusion pressure, ram speed, surface finish, aged strength and anodised brightness when extruded at $450^\circ C$.

2.2.2. Silicon (Si)

Silicon combines with iron in preference to magnesium⁽²⁴⁾ to form an $AlFeSi$ phase. The silicon available for magnesium silicide is therefore given by⁽²⁸⁾: $\%Si - 0.25(\%Fe + \%Mn)$.

An excess of silicon over that required for magnesium silicide and iron phases will increase the strength^(22,24,25,28,30,38,39) but decrease the extrudability^(22,28,30,38). The influence of silicon on extrusion speed, for alloys with constant levels of 0.4% and 0.5% Mg, has been shown⁽²⁰⁾ to decrease the ram speed by 0.4m/min per 0.01wt%Si increase.

Notch sensitivity⁽²³⁾ and brittleness^(24,29,30) are reported to increase with excess silicon. The loss of ductility is considered to

be due to⁽²⁰⁾ Si and Mg_2Si precipitates in the grain boundary. Loss of toughness^(25,30,38) can be controlled, however, with minor additions of manganese and/or chromium^(23,29,38,40).

The surface finish of extrudates produced with excess silicon alloys may display deleterious dark spots⁽³⁰⁾ which are caused by silicon precipitates.

2.2.3. Magnesium (Mg)

An excess of magnesium in the AA6XXX series alloys is much more detrimental to the extrudability^(30,38) than an equal amount of excess silicon, and contributes little to strength⁽³⁸⁾. For a constant 0.43% and 0.5% silicon level and magnesium content <0.55%, extrusion speed is decreased by 0.4m/min per 0.01wt%Mg increase⁽²⁰⁾. At levels of magnesium >0.55% the speed decrease is more rapid. This is because there is a change in the mechanism of tear initiation⁽²⁰⁾. This phenomenon is discussed later.

Corrosion resistance is reported⁽³⁹⁾ to improve with excess magnesium.

2.2.4. Iron (Fe)

Iron is present in AA6XXX series alloys as a controlled impurity⁽²⁷⁾, hence, even the most simple of AA6XXX alloys may be considered as a quaternary alloy of the Al-Fe-Mg-Si system.

Silicon combines with iron in preference to magnesium⁽³⁵⁾ to form an AlFeSi phase. An increase in the iron content can therefore decrease the strength of the alloy by removing excess silicon (especially when there is inadequate silicon remaining to combine with magnesium). This is often a cause of low mechanical properties in AA6063-T5⁽²⁸⁾. However, a secondary effect of the iron phase is to refine the magnesium silicide precipitate thus producing an increase in the T6 strength⁽²⁴⁾, this being more so in higher excess silicon alloys.

The βAlFeSi phase, which is present in as-cast billets, has an adverse effect on the hot working characteristics⁽⁴²⁾. To permit higher extrusion speeds⁽²⁶⁾, without the appearance of intermittent finite white fleck surface tears⁽³⁶⁾, an homogenisation treatment must be used to refine and transform the βAlFeSi into the preferred αAlFeSi . A typical industrial homogenisation treatment⁽²⁹⁾ is 6 hours at 565°C.

Iron is also reported to cause a reduction in corrosion resistance⁽³⁹⁾ and improve toughness⁽⁴⁰⁾.

2.2.5. Copper (Cu)

Europe does not approve the additions of copper to AA6XXX series alloys due to the alleged poor corrosion resistance^(22,35,43). This is contrary to the findings of some workers^(22,38,39) who show that, in balanced alloys, small additions of copper have no detrimental effect. However, in alloys with excess silicon, low levels of copper have been

found⁽³⁸⁾ to significantly decrease corrosion resistance under severe conditions.

Small additions of copper enter aluminium solid solution and iron rich phases⁽³⁸⁾, with copper phases only occurring at levels greater than 1%⁽¹³⁾.

Minor copper additions improve strength^(22,24,25,35,39,44) by entering solid solution and also by refining the Mg_2Si precipitate^(24,25,35,38). The refining effect of copper is additive to that produced by excess silicon⁽²⁴⁾.

A low level of copper has no noticeable effect on toughness, however, it does to some extent inhibit grain boundary precipitation and influences deformation within the grains⁽²⁴⁾.

Copper is added to alloys AA6063 and AA6463 to enhance brightness on bright dipping and anodising⁽²⁸⁾.

2.2.6. Manganese (Mn), Chromium (Cr) and Zirconium (Zr)

The three minor additions Mn, Cr, and Zr are added to improve toughness^(22,29,45) by forming fine, stable incoherent dispersoids during homogenisation^(23,35). However, these additions increase quench sensitivity^(35,38) by providing nucleation sites for the Mg_2Si .

Increasing the Mn and Cr beyond that required for adequate toughness

will further decrease extrudability^(22,28,35). Mn is preferred over Cr because it is far less detrimental to extrusion speed and finish⁽³⁸⁾. However, it is more harmful to quench sensitivity⁽⁴⁴⁾.

Recrystallisation is inhibited with the addition of Mn and Cr by forming intermetallic particles which are capable of pinning grain boundaries.

Small percentages of Mn reduce pick-up⁽⁴⁶⁾ and improve surface finish by promoting a fine distribution of Mg_2Si . The introduction of Mn also accelerates the transformation of the non-equilibrium ' β ' to the equilibrium ' α ' iron phase, during homogenisation⁽³³⁾.

2.2.7. Other Minor Additions

Vandium (V) may be added to promote a fine grain size on casting⁽⁴⁷⁾, this leads to improved strength, toughness and formability.

Lead (Pb) is sometimes added to improve machinability.

Zinc (Zn) is present as an impurity⁽²⁷⁾. Zn increases strength slightly⁽³⁹⁾ by entering solid solution.

Titanium (Ti) and Boron (B) control grain size^(27,39) and increase the solidification rate⁽³⁶⁾.

2.3. Phase Transformations in Al-Mg-Si Alloys

The majority of phase transformation studies in this alloy have been undertaken on so-called 'balanced alloys' where solute additions are such that the alloys may be considered as a pseudo-binary (Al-Mg₂Si) system. The decomposition of supersaturated solid solution in Al-Mg-Si alloys has been studied extensively by X-ray⁽¹⁰²⁻¹⁰⁴⁾ and transmission electron microscopy^(31,99,105,106) investigations. The precipitation sequence in this alloy has been proposed⁽¹⁰⁷⁾ as:-

Needle shaped zones $\rightarrow$ Needle shaped zones $\rightarrow \beta'$ $\rightarrow \beta$ (Mg₂Si)
 along $\langle 100 \rangle_{Al}$ with internal order

The β' rods were determined⁽¹⁰⁶⁾ to be hexagonal ($a=7.05 \pm 0.5$, $c=4.0.5 \pm 0.05$) and form in the cube direction. The equilibrium β (Mg₂Si) phase forms as platelets with an f.c.c. (CaF₂) structure ($a=6.39$).

In AA6063 alloy the second phase dissolution is of considerable importance, as its overall process economics relies on attaining sufficient dissolution during hot extrusion. This eliminates (with a suitable quench rate immediately after extrusion) the need for a separate solution treatment before artificial ageing treatments. Therefore, the homogenisation treatment must render the Mg₂Si in a form, suitable for rapid dissolution during the short excursion above the solvus temperature, when extruded. Under common (industrial) practices the short homogenisation treatments and cooling conditions produce β' rods which dissolve more readily than equilibrium β , during billet preheat and subsequent extrusion at 450°C to 480°C.

In some recent work⁽³¹⁾ the use of step cooling was seen to increase dissolution of β' rods by the precipitation of α -phase on the β' interface which on reheating was thought to provide local sink sites for the solute fluxes driving the dissolution process. However, these experimental findings were confined to small sample sizes and carefully controlled furnace temperatures, which industrially would be considered excellent, if not impossible, on large scale production. Also there was no comparison made between step-cooling and normal continuous cooling treatments during and after extrusion in order to justify the additional thermal processing.

Another important phase transformation to be made, especially when considering surface finish, is that of the iron phase. This is discussed in the following section.

2.4. Homogenisation

After a billet is cast it must be homogenised to improve extrudability and the final mechanical properties of the extrudate. In this treatment the billet is held at a sufficiently high temperature to achieve complete, or near complete, solution of the soluble constituents and to disperse them uniformly throughout the cross section. Homogenisation may also serve to transform insoluble phases into a less detrimental form.

The homogenisation time and temperature in AA6XXX series alloys must be sufficient to reduce the microsegregation and coring which occurs with rapid solidification in D.C. casting. Homogenisation will

improve extrudability^(28, 29) by converting the AlFeSi phase from the β form found in as-cast billets to the less detrimental α phase. The β to α transformation involves loss of silicon from β into the matrix, which not only renders the α -iron phase less brittle but also allows more silicon to be precipitated as Mg_2Si . The iron phase transformation is complete after 6 hours at $565^\circ C$ ⁽²⁹⁾. However, the introduction of small amounts of Mn shifts the α AlFeSi phase field in the direction of a lower Fe/Si ratio, thereby accelerating the β to α transformation. The initial transformation occurs without any shape change, but continued thermal treatment results in a gradual rounding of the precipitate.

The Mg_2Si enters the solid solution rapidly at temperatures in excess of $465^\circ C$ for AA6063 and generally requires as little as 2 hours at this temperature. Therefore, hold time and temperature of the homogenisation treatment for the AA6XXX series alloys is determined by the iron phase, as opposed to the magnesium silicide constituent.

After adequate homogenisation treatment, the critical factor affecting the condition of the homogenised billet is the cooling rate⁽³³⁾. A slow cooling rate promotes coarse precipitates of Mg_2Si , especially in the temperature range of between $316^\circ C$ to $417^\circ C$ ⁽²⁷⁾. It has been shown⁽⁴⁹⁾ that a cooling rate of $38^\circ C/hr$, in AA6063, results in an average precipitate size of approximately $2\mu m$. To redissolve precipitates of this size, extended soak times, prior to extrusion, would be necessary. This is often impractical, and when low extrusion temperatures are employed, much of the precipitate would remain undissolved and add nothing to strength. The excessive heterogeneity

would also cause surface pick-up and surface microcracking⁽³²⁾.

A fast cooling rate after homogenisation will retain most of the magnesium and silicon in solution and produce small and uniformly dispersed precipitates. For example, a cooling rate of 315°C/hr is reported⁽⁴⁹⁾ to give a precipitate size of approximately 0.25µm. Such fine precipitates are readily dissolved during extrusion, without the need for a pre-solution soak. This is essential to enable full age hardened mechanical properties to be obtained from the extrudate, without the need for a separate solution treatment between extrusion and artificial ageing.

Homogenisation treatments are also determined by the required surface finish of the end product. For example, a low homogenisation temperature of 530°C to 550°C is preferred for caustic etch and anodised finishes, to give a soft matt appearance. This finish is enhanced in alloys with a high iron content⁽²⁸⁾. High homogenisation temperatures of 580°C to 590°C increases the brightness of bright dipped and anodised finishes⁽²⁸⁾, particularly in AA6463 and AA6763.

2.5. Hot Working

2.5.1. Empirical Relationships

Several authors^(5-9,11-14) have shown that an equation, originally proposed for creep, is applicable to steady state hot working data. This equation is most commonly quoted in a rearranged form to include

the Zener-Hollomon⁽⁵⁷⁾ parameter:-

$$Z = \dot{\epsilon} \exp(\Delta H/RT) = A(\sinh(\alpha\sigma))^n \quad -(2.1.)$$

where: $\dot{\epsilon}$ = effective strain-rate (s^{-1})
 σ = effective stress (MNm^{-2})
 T = absolute temperature (K)
 ΔH = activation energy
 α, n, A = material constants
 R = universal gas constant.

The Arrhenius term indicates that hot working is a thermally activated process⁽⁸⁸⁾ with ΔH equal to the activation energy of the rate controlling mechanism. For high stacking fault energy materials such as aluminium and its alloys, the activation energy for hot working is similar to that for bulk self diffusion, which is attributed to the ease of dislocation climb in this type of material.

The use of the constitutive Equation (2.1.) is justified at high temperature deformation conditions because the operation of dynamic recovery and repolygonisation renders the flow stress independent of strain.

2.5.2. Mechanical Equation of State

A constitutive law such as the Zener-Hollomon equation, adequately describes and predicts the deformation response of aluminium to various imposed deformation rates and temperature exposures, in the hot working regime. However, an increasingly important use of constitutive equation is to relate deformation behaviour to varieties of metallurgical micro-structure. A fundamental trait of the deformation behaviour of a metal is that its current properties depend entirely upon its current metallurgical structure. Furthermore,

changes in those properties result from structural changes, which can occur as the property is being tested, and also when unconstrained and static. Because plastic deformation of crystalline solids is accomplished almost exclusively by mobile structural imperfections, the dislocation structure occupies a special place in relation to mechanical properties. Therefore the plastic behaviour of a metal can be divided into two mechanistic steps: the flow stress (σ) depends on strain-rate ($\dot{\epsilon}$), temperature (T) and the current structure (ρ).

$$\sigma = \sigma(\rho; \dot{\epsilon}, T) \quad \text{---(2.1.a)}$$

and the structure evolves with strain:-

$$\frac{d\rho}{d\epsilon} = (\rho) \quad \text{---(2.1.b)}$$

For simplicity, it is assumed, in the above relationships, that the dislocation structure can be represented by a single parameter ' ρ ' which reflects some appropriate average of the dislocation distribution. Equations (2.1.a) and (2.1.b) also imply that, at a given ρ , the flow stress depends on $\dot{\epsilon}$ and T , but the evolution of ' ρ ' does not. Such rate-independence is implicit in phenomenological plasticity.

The use of Equations (2.1.a,b) requires a suitable back extrapolation method to be found which eliminates all effects of on-going structure change⁽⁹⁴⁾ which is not necessarily possible. The flow stress may also depend on the current rate of change of the dislocation structure⁽⁹⁵⁾. Therefore, separation of constant-structure kinetics from evolution is not clear-cut. Mecking and Kocks⁽⁹⁶⁾ generalised Equations (2.1.a,b) in two respects: to allow both flow stress and the

rate of evolution of the sub-structure with strain, to depend on $\dot{\epsilon}$ and T , and that it requires two parameters, ρ_1 and ρ_2 to represent the macroscopic effects of the sub-structure adequately.

$$\sigma = \sigma (\rho_1, \rho_2, \dot{\epsilon}, T) \quad -(2.1.c)$$

$$\frac{d\rho_1}{d\epsilon} = (\rho_1, \rho_2, \dot{\epsilon}, T) \quad -(2.1.d)$$

$$\frac{d\rho_2}{d\epsilon} = (\rho_1, \rho_2, \dot{\epsilon}, T) \quad -(2.1.e)$$

The introduction of two state parameters requires two evolution equations to be identified, whose integration may present some difficulty.

Anand⁽⁹⁰⁾ proposed a simplified approach to the notion of a mechanical state equation by formulating a phenomenological theory with a single, non-zero, positive valued scalar (s) to represent the internal variable:-

$$\sigma = \sigma (s, \dot{\epsilon}, T) \quad -(2.1.f)$$

$$\dot{s} = s (s, \dot{\epsilon}, T) \quad -(2.1.g)$$

Such a simplified approach may be justified when the load application is uniaxial and of constant sign⁽⁹⁷⁾.

The Kinetic equation used by Anand was that proposed by Lee and Zaverl⁽⁹²⁾

$$\sigma = s \cdot \left(\frac{\dot{\epsilon}}{A} \exp (\Delta H/RT) \right)^m \quad -(2.1.h)$$

(The symbols take their usual meaning).

The evolution equation for 's' was then formulated (ignoring static softening):

$$\frac{ds}{d\epsilon} = h_0 \cdot \left(1 - \frac{s}{s^s}\right) \dot{\epsilon} \quad -(2.1.i)$$

where: $s^s = \hat{s}(\dot{\epsilon}, \exp(\Delta H/RT))^n$

The important qualitative features embodied in Equation (2.1.i) are that the effective hardening modulus 'h' is composed of a constant hardening rate 'h₀' and a variable softening rate 'h₀s/s^s. Further, s^s is the saturation value for 's' at a given $\dot{\epsilon}$, and T pair.

A successful application of this type of approach was made by Sample and Lalli⁽⁹⁸⁾ for compression of commercial purity aluminium at hot working temperatures. They presented a relationship between the developed micro-hardness and the state parameter, which displayed good predictive capabilities. However, it is likely that such a relationship would be difficult to determine in alloys where contributions to hardness originate from numerous sources (e.g. solution hardening, and grain size effects).

Despite the limitation of the Anand model, the approach is worth pursuing, as an initial study, to establish a mechanical state equation.

2.5.3. Restoration Processes

When a material is cold worked (i.e. at temperatures less than $0.1T_m$; where T_m is the melting point in Kelvin), it will 'work harden'. This phenomenon is caused by the interaction of dislocations which are generated during deformation.

At hot working temperatures ($>0.6T_m$), the effect of work hardening is reduced by dynamic softening processes. It is these restoration processes that allow large strains, as found in extrusion, to be obtained by rendering the material strain independent in the plastic region. There are two dynamic softening mechanisms that can occur, which can act solely or in combination. These mechanisms are termed 'Recrystallisation' and 'Recovery' which have received extensive reviews(2-9,11-14,50-53) and so will only be summarised here.

The restoration processes occur during, and subsequent to, hot working and are termed 'dynamic' and 'static' respectively.

'Dynamic recovery' involves the annihilation of pairs of dislocations. In high stacking fault energy (SFE) materials, thermal activation assists the dislocations to cross-slip and climb in order to achieve extinction. It is dynamic recovery that is responsible for the formation of sub-grains with low angle boundaries. However, in materials with low SFE, local differences in the dislocation density are sufficient to permit the nucleation of recrystallisation during deformation (i.e. 'dynamic recrystallisation'), which leads to the elimination of large numbers of dislocations by the migration of high

angle boundaries.

It is well established that dynamic recovery operates during hot working of aluminium alloys, leading to the replacement of the original grains by well defined sub-grains. The initial stages of the sub-structure development are described by Luton and Sellars⁽⁵⁴⁾, and the steady state regime has been well documented^(2-9,11-14,50). When the steady state regime of the deformation has been established, the flow stress remains constant and the sub-grains maintain an equilibrium size and shape by continuously reforming⁽⁵⁰⁾. This process has been coined 'repolygonisation'.

The size of the sub-grains increase with temperature and with a decrease in strain-rate, and has been shown^(7,11,14,50) to be related to the Zener-Hollomon parameter 'Z' by a relationship of the form:-

$$d^{-m} = a + b \ln Z \quad \quad \quad -(2.2.)$$

The dependence of the sub-grain size on the temperature-compensated strain-rate and their relationship to room temperature mechanical properties⁽⁵⁰⁾ is an important consideration in extrusion, since it should enable the final properties of the extrudate to be related to the process variables.

Dynamic recrystallisation is not generally observed in aluminium alloys, though it has been reported for some Al-Mg alloys^(7,14,55).

As soon as deformation is complete, the release of stored energy in

the form of the sub-structure introduced during working, acts as the driving force for static restoration processes to be initiated. Static recovery involves the annihilation of dislocations, in individual events, while in static recrystallisation, the dislocations are simultaneously eliminated in large numbers as a result of the motion of high angle boundaries. Recovery is brought about by changes in the number and distribution of point defects and dislocations, while recrystallisation involves replacement of the worked substructure by a new dislocation-free one. Dynamic recrystallisation can also provide the nuclei for metadynamic recrystallisation⁽⁵²⁾, which requires no incubation period, and can also take place concurrently with static recovery.

If a metal reduces its stored energy through static recovery, then recrystallisation is less likely to occur. Therefore, in materials where dislocation climb and polygonisation are relatively easy, the possibility of recrystallisation is reduced⁽⁵⁶⁾. In metals with a distribution of large second phase particles an acceleration of the nucleation process may occur, thereby enhancing static recrystallisation.

2.5.4. Structural Aspects (Extrusion)

Extrusion produces a marked directionality in the structure, in metals that dynamically recover, due to the very high strains that are encountered. The original grains are elongated with an interior of low misorientation sub-grains and the precipitates are aligned in the extrusion direction. However, the strain produced is not uniform

across the extrude, with surface layers undergoing considerably more deformation, and at a higher strain-rate in passing the face of the dead metal zone, and along the die lands. The higher strains in the surface layers may lead to a higher dislocation density, which can often promote static recrystallisation. This leads to an annulus of recrystallised grains in the extrudate.

The grains' directionality in the extrusion of aluminium, due to preferred orientation is [111] and [100]; 75% being of the type [111]⁽⁵⁸⁾. The texture produced increases the strength of the extrudate in uniaxial tension from the [111] directional grains. However, the strength of the extrudate is decreased when a recrystallised annulus occurs.

2.6. Hot Working by Extrusion of Al-Mg-Si

The general principles of hot working were described in the previous section and are the subject of a number of review papers⁽⁵⁰⁻⁵²⁾. In the following section, those aspects specific to the hot working of Al-Mg-Si alloys will be reviewed.

The Al-Mg-Si alloy system, more than any other, is fabricated by extrusion, in the hot working regime. Therefore, it is not surprising that the vast majority of studies have focussed on this process.

A great proportion of extruded shapes in Al-Mg-Si alloys are made by the conventional process of (unlubricated) hot direct extrusion through flat faced dies; the billets being preheated to a temperature

in the range of 420°C to 500°C. Under these conditions, and with a suitable pre-extrusion homogenisation treatment, the AA6XXX series alloys can be extruded at high speeds, with good surface quality, giving high rates of production.

In dealing with AA6XXX extrusion alloys it is most common to consider the extrusion process as a combined hot-forming and solution heat treatment, to be followed by a suitable quench which prepares the material for subsequent artificial ageing.

A metallurgical structure most favourable for high rate extrusion is one with the lowest feasible amount of magnesium in solid solution because of its adverse effect on flow stress⁽²⁸⁾. Conversely, the highest strength properties are obtained when the maximum possible of Mg and Si are in solution as the extrudate emerges from the die. Therefore, the process becomes one of compromise and increased importance on the homogenisation practices (these were discussed in section 2.4.).

Dilute alloys such as AA6063 have a low quench sensitivity, which permits air quenching of the emerging extrudate to retain the Mg and Si in solution. This generally means that the structure will be completely recrystallised. The statically formed grains often appear to have similar orientations, reflecting the fibre textures produced during extrusion⁽⁵³⁾.

When static recrystallisation occurs, large numbers of dislocations are eliminated simultaneously as a result of the motion of high angled boundaries, from a point of nucleation.

If disperoid-forming elements are added, recrystallisation will tend to be inhibited. Static recrystallisation can also be reduced by water quenching immediately as the extrudate emerges from the die. In these cases the recrystallised grains are replaced by a fibrous hot worked structure consisting of fine equiaxed sub-grains with well defined boundary walls. These are formed by the restoration mechanism, 'dynamic recovery' which is operative in Al-Mg-Si alloys during steady state hot extrusion. The thermally activated cross-slip and climb of dislocations, leads to the formation of low misorientation ($1-5^{\circ}\text{C}$) sub-grains within the original deformed grain, by the summation of individual (generation and annihilation) events. When the sub-gains are retained in the extrusion product, their mean size is determined by the temperature and strain-rate conditions utilised during extrusion: an increase in temperature and/or a decrease in strain-rate leads to a larger sub-grain diameter^(50,51). This type of microstructure adds an increment of strength over that possible simply by heat treatment. This is thought to arise because the sub-boundaries oppose dislocation motion⁽⁵³⁾.

The retention of the hot worked structure is more difficult at the surface of the extrudate due to the increased deformation received (and therefore temperature increase) and the greater number of nucleation sites available for recrystallisation.

Billet pre-heating to extrusion temperature is also an important aspect of the complete extrusion process. Similar to cooling from the homogenisation temperature, Mg_2Si precipitation is also possible during billet pre-heat. Therefore, in order to retain the beneficial

features achieved from the homogenisation treatment, rapid pre-heating is required. This can be achieved by using induction heating as opposed to the lengthy gas furnace heating.

The attainment of the maximum productivity, with best possible surface quality and mechanical properties, therefore requires careful control of all aspects of the operation from billet casting, through homogenisation, reheating, extrusion and all post extrusion treatments.

2.7. Metal Flow During Extrusion

The greatest contribution to the knowledge of the metal flow in extrusion is due to research undertaken by Pearson and Parkin⁽¹⁾. Their method of investigation was a split billet technique. This involved splitting a cylindrical billet along its central axis, machining a grid onto one of the cut faces, and then partially extruding the re-matched billet, axisymmetrically. The billet was then removed from the container to be subsequently prised open to reveal the distorted grid pattern. This method does not affect the flow during extrusion because there are no shear stresses in the longitudinal plane.

The flow patterns of materials during extrusion have been divided into four classes; S, A, B and C. These are shown schematically in Figure (2.2.). The differences in the flow pattern is caused primarily by the frictional resistance at the billet/container interface. Flow stress variations across the billet, and temperature differential between the

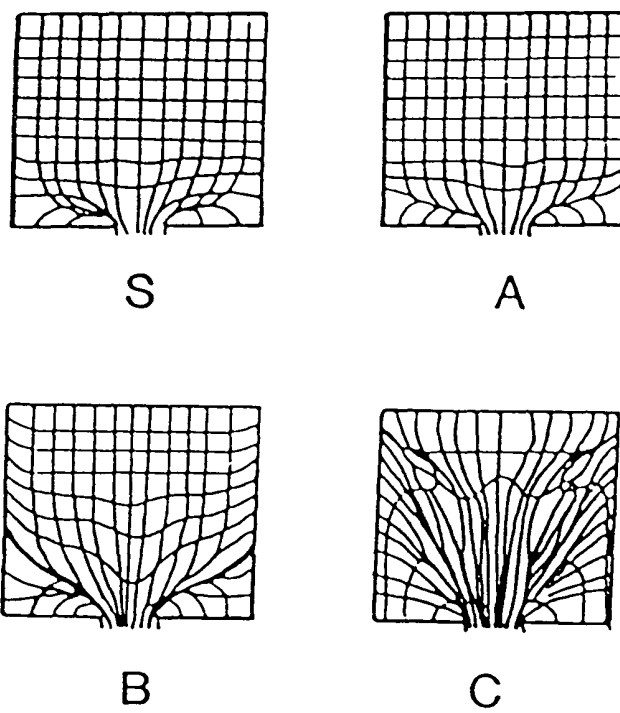


Figure 2.2. Types of Extrusion Flow Pattern

billet and container will also determine the pattern of flow.

Flow Pattern 'S'

This pattern represents the maximum possible uniformity of flow in the container. The major portion of the non-extruded billet is moved as a rigid undeformed body. The billet remains undeformed until plastic flow occurs in the deformation zone positioned directly in front of the die. A small dead metal zone is formed on the die face adjacent to the die throat because the metal cannot flow around the sharp corner between the container and the die.

This type of flow can only occur when there is no frictional constraint. This is obviously a hypothetical case, though when a very effective lubrication is used and in indirect extrusion with good die lubrication, this flow pattern can be approached.

The surface of the extrudate in this type of flow is partly generated from the surface of the billet, hence, if the billet is not scalped prior to extrusion or extruded leaving a shell, the surface finish of the extrudate will be inferior due to the surface oxide and inverse segregation of the billet, flowing into the product surface.

Flow Pattern 'A'

When there is very little friction between billet and container but significant friction at the surface of the die, a flow pattern 'A' is obtained. The friction at the die face results in a retardation of the radial flow of the peripheral zones and increases the amount of

shearing in this region. This leads to a slightly larger dead metal zone and wider deformation zone than that found in flow type 'S', however, the flow is still relatively uniform. Flow patterns of this type may be found when extruding lubricated soft alloys such as lead, tin and alpha brasses.

Flow Pattern 'B'

This pattern occurs if there is friction at both the container wall and at the surface of the die. The retarding effects of the friction on the metal flow produces a large dead metal zone.

At the start of the extrusion the shear deformation is mainly in the peripheral regions of the billet. However, towards the end of the extrusion cycle it extends towards the centre, which increases the danger of the material from the billet surface flowing into the extrudate. The impurities or lubricant from the billet surface may then cause blistering.

The 'B' type flow pattern is observed in most aluminium alloys.

Flow Pattern 'C'

The 'C' type flow pattern is observed in hot extrusion when friction is high. This type of flow is enhanced in the materials where the flow stress of the cooler peripheral regions is considerably higher than that of the centre. The dead metal zone formed in this type of flow is very large and extends from the front of the billet to the back.

Extrusion defects known as 'pipe' and 'funnel formation' can occur in materials exhibiting the flow pattern 'C'. Typical alloys in this class are ($\alpha+\beta$) brasses.

2.7.1. Extrusion Defects Produced by Undesirable Flow

Extrusion defects may be visible, visible after treatment, or invisible (internal). Bryant⁽³⁴⁾ classified defects according to their appearance and studied the extrusion process from preparing the billet, to cooling the extrudate to room temperature, in order to study the metallurgical effects of each stage on the extrusion defect. However, defects produced due to undesirable flow are generally only visible on cutting the extrude to examine the internal structure.

2.7.2. Funnel Formation

The effect of friction at the container wall and die face is to cause the material flow at the centre of the billet to be more rapid than at the edge. Therefore, as the end of the billet approaches the die, a funnel shaped hollow forms in the discard. The funnel can often extend into the extrudate when an insufficient discard length is left.

2.7.3. Pipe Defect

The pipe defect occurs at the periphery of the dummy block when the

surface of the billet, which contains oxides, turns into the more rapidly flowing material in the centre of the billet. The metal, flowing into the extrudate is unable to weld to the adjacent extrudate metal, due to oxides and other surface contaminants, thereby resulting in a product with an outer and inner core. The defect is usually found in the remaining 30% of the extrudate and in materials exhibiting a 'C' type flow pattern.

There are three main techniques that may be employed to reduce pipe formation. The first method is 'extrusion with a shell'. Here a dummy block, slightly smaller in diameter than the billet, is used to leave a shell of the oxide-clad, billet surface in the container. The shell can then be pushed out after extrusion. Care must be taken, however, in choosing the dummy block diameter because if it is too small (i.e. a thick shell), back extrusion may occur. Machining the billet prior to extrusion will also reduce pipe formation by removing the impurities and oxides from the billet surface. This method is limited, of course, by the oxidation of the material to be extruded during storage and on heating to extrusion temperature.

The third technique to eliminate this defect is to 'extrude with a long discard'. A clear disadvantage in employing this technique is the high proportion of material scrap, and its associated recycling costs. Moreover, in some metals, such as titanium this cost is not economically viable. Methods which can reduce the friction from the container walls and the die face will reduce both pipe and funnel formation.

2.8. Surface Generation and the Formation of Defects

Regardless of the flow pattern type, die design and lubrication conditions, the generation of an extrudate surface involves, to some extent, the production of a virgin surface by shearing from within the billet. However, the surface formation mechanism in non-lubricated, direct extrusion, through a flat faced die, is quite different from that of lubricated extrusion through a conical die. In the latter case the billet surface slides along the container wall and die face, during which time it elongates before leaving the die orifice to form part of the product surface. (The percentage of newly generated surface increases with reduction ratio.) In non-lubricated, direct extrusion, through a flat faced die, sticking friction normally exists at the interface between the tooling and billet. This results in the formation of a dead metal zone ahead of the die face. (The metal flow is of type B or C.) The consequence of this pattern of metal flow is that the extrudate surface is generated totally by shearing inside the body of the billet. This particular surface has been termed⁽⁵⁹⁾ 'frisch entstandere oberflächen' (freshly formed surfaces) to distinguish them from surfaces formed in lubricated extrusion consisting partly of the billet surface.

Valberg⁽⁶⁰⁾ in a recent study on the surface formation mechanism in the direct extrusion of aluminium alloys, reported that the material forming the extrudate surface originated from a region near to the dead metal zone. Valberg concluded that the surface formation mechanism in direct extrusion could be described as a cladding process where metal from the front end of the billet, stretches out, during the extrusion cycle, to clad the surface of the resulting profile.

However, such a conclusion was unavoidable as the investigative method used in his experiments was, in fact, a commonly used cladding technique. This involved placing a disc of material ahead of the extrusion billet such that when extruded, the material from the disc stretches out over the extrudate formed from the billet material. Unfortunately there is little other published work in the technical literature on the topic of surface generation. This is clearly a serious omission as it is the surface generation mechanisms that will govern surface defect formation.

The incidence of surface defects is of considerable concern to the extruder as the vast majority of product demand lies in the architectural and decorative trim market. The surface defects which may be generated during the extrusion process have been divided into four main categories⁽³⁴⁾; (a) cracks and tears, (b) pick-up and die lines, (c) blisters and (d) streaking.

2.8.1. Cracks and Tears

Cracks and tears involve two main mechanisms and give rise to the defects which are commonly termed speed cracking and incipient melting. Speed cracking is a result of localised tensile failure at the surface of the extrudate caused by a reduction in the flow stress. Moreover, there is a general loss in ductility of the material, when the hydrostatic component of stress is released, as the metal enters the die throat. This combination of low flow stress and limited ductility allow the friction forces in the die land area to cause surface cracking by a tensile mechanism. It has been shown^(7,9,11)

that this is a result of thermal instability and is related to the initial billet temperature and to the strain-rate. The conditions for acceptable surface finish may thus be expressed in terms of the Zener-Hollomon parameter such that:-

$$\ln Z < aT^{-b} \quad \text{---(2.3.)}$$

if the surface is to have an acceptable appearance.

In Equation (2.3.), 'Z' is the Zener-Hollomon parameter as defined by Equation (2.1.) and 'T' is the temperature; 'a' and 'b' are constants. Speed cracking equations have been quantified in these terms for various alloys such as AA2014, AA5456 and AA7075⁽⁶¹⁻⁶³⁾.

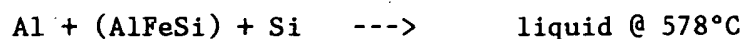
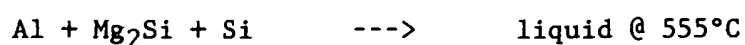
When incipient melting occurs it is sometimes termed the 'fir tree defect' because of its appearance. The defect is most common in aerospace and structural alloys which have low melting point phases. The defect is a result of the surface temperature exceeding the liquidus or second phase melting temperature, and hence its incidence may also be quantified by consideration of the Zener-Hollomon parameter yielding an equation similar to Equation (2.3.).

The AA7XXX alloys are especially susceptible to hot shortness because of the low melting S-phase and it has been shown^(8,64) that the maximum temperature permissible during the extrusion process is as low as 480°C. Because of the increased deformation suffered by the material feeding the extrudate surface, the temperature rise at this location may be sufficient to exceed the incipient melting temperature, thereby initiating the crack at the extrudate surface.

Moreover, when sections are extruded, Lange reports⁽⁶⁵⁾ that, based on previous experience, the temperature rise at the edges of a section will be twice that at the mid-point of the sides. This explains the commonly observed crack initiation at the corner locations, of the extrusion product.

The alloy under investigation in the present work, AA6063, has a very high extrudability and generally results in a solid product even at temperatures close to the matrix melting temperature. However, two types of crack initiation have been identified⁽²⁰⁾ in AA6XXX series alloys. The first is observed when the alloy content is low and was thought to be due to tearing in the aluminium matrix. At higher alloy contents (>0.55%) the cracks appeared to initiate via a second mechanism. In this case, the presence of large precipitates near the extrudate surface created a weakened volume which could be opened up by the stresses at the die, leaving a void in the subcutaneous structure, which subsequently leads to cracking.

Incipient melting of the low melting point eutectics are avoided in dilute alloys such as AA6063 when correct homogenisation and cooling practices are utilised. However, in more highly alloyed AA6XXX series alloys and where inadequate homogenisation practices are performed, two possible eutectics have been indentified⁽²⁹⁾ which reduce the maximum permissible extrusion temperature. These possibilities are:-



Attempts at avoiding melting by using lower ram speeds for common alloy extrusions severely reduces production and so emphasises the importance in determining the optimum homogenisation treatment.

Another type of defect under this heading is 'surface tearing'. This defect is most severe when extruding through a new die⁽⁶⁶⁾. The tears are caused by the unstable nature of the aluminium adhesion on the die land area. When a new die is commissioned, adhesion of the extrudate surface is intermittent both across and along the die land area. This results in areas of the extrudate material being torn from the adjacent flowing surface. The severity of this type of tearing decreases with the initial billets processed, as a stable layer of aluminium is deposited on the die land area. The coated die land thus eliminates tearing but can be one of the causes^(34,66-68) of the defects which fall under the category 'pick-up and die lines'.

2.8.2. Pick-Up and Die Lines

Probably the principal defects occurring during extrusion of aluminium alloys fall under this heading, and are most serious in those alloys where cosmetic appearance is a prime requisite. Despite the commercial importance of such defects, especially in the common alloys such as AA6063, there remains much conjecture over their cause.

Present evidence suggests that during extrusion, a thin film of extrudate material adheres to the exit side of the die land area and in so doing collects coarse precipitates and oxide particles which shear the extrudate surface. The defects thus appear as longitudinal

drag lines which may be 0.25mm-0.1mm deep and vary in length between 3mm and 12mm; they often terminate in flecks of aluminium debris adhered to the extrudate surface which is referred to as the 'pick-up deposit'. The defect is generally thought to be caused by the following: inclusions in the cast billet⁽³⁴⁾, inadequate homogenisation treatment^(32,34), and die deflection⁽⁷¹⁾; the defect also appears to be temperature sensitive. Remedies to reduce the occurrence of this defect has concentrated on many aspects of the extrusion process, but have tended to target on die modifications such as converging die lands⁽⁷¹⁾, so that under extrusion load the lands become parallel. The use of nitrided or chromised⁽⁷¹⁾ tool steel, or the use of die materials less prone to aluminium build-up such as Zr_2O_3 ⁽⁶⁹⁾ and Syalon⁽⁷⁰⁾, are all reported to reduce pick-up.

Among the more novel approaches to tackling the pick-up problem is the insertion of a leader plate, 1mm-2mm thick, ahead of the billet⁽⁶⁸⁾ with the addition of 0.01%-0.5% of Boron to the extrusion billet's composition. (The introduction of a fine distribution of small, and hard boride particles effectively maintained a clean die land area by a scouring action from the leader plate.)

Another approach is the employment of an inert atmosphere to prevent oxidation of the emerging extrudate⁽⁷²⁾ and the use of cooled die tooling⁽⁷³⁾. The use of die cooling is not a new idea; the first patent was registered in the USA in 1866 for cooling the die with water. However, a more recent German patent uses liquid nitrogen to both conduct heat away from the die and prevent oxide formation on the exit side of the die by displacing the atmospheric oxygen with the evaporating nitrogen⁽⁷⁴⁾. The surface of the extrudate under such

processing conditions was reported to be relatively free of pick-up particles and possessing shallower die lines.

Die lines are also believed to be caused by the interaction of the billet material with the die land area^(34,66,68). Die lines appear on all extruded products, but are generally tolerated in high strength aluminium alloys as they do not appear to affect mechanical properties. However, for AA6XXX series alloys, where aesthetic quality is the most commercial aspect, their occurrence is most undesirable. In an attempt to conceal the die lines, a trend for a high degree of mattness or diffuse appearance, by etching, has been established in the market⁽⁷⁵⁾. Endeavours to reduce or eliminate die lines from the extrusion process have included machining (and polishing) the billet, and more practically, the use of highly polished dies and the introduction of lubricants (as an uninterrupted separating film). However, even when these precautions are taken, the die lines still occur, but are fewer in number and appear finer. In this event they are termed 'micro-die lines'⁽⁶⁶⁾ and are thought to be a direct result of the deformation mechanism operative during extrusion. To date, there is no published research on the formation mechanism of the micro-die lines.

2.8.3. Blisters

The blistering defect may occur when there is an introduction of lubricants and when the metal flow is of pattern type 'S', 'A' or 'B'. This type of defect is caused by the flow of lubricant traces, from the billet surface, along the dead metal zone and into the

subcutaneous layers of the extrudate, and as the temperature increases during the process, turns gaseous thus blistering the surface. The defect may also be caused by oxides and other impurities from the billet surface behaving in a similar manner by trapping an air pocket below the surface. Hence it is unusual to extrude aluminium alloys using flat faced dies and in the case of aerospace alloys, to scalp the billet prior to extrusion, or extrude with a remnant shell.

2.8.4. Streaking (And Other Post Extrusion Defects)

Surface defects can still occur, even when an apparently excellent extrudate surface finish is obtained as it leaves the die. One of the most common defects in this category, appearing on AA6XXX series extrudates is streaking. This occurs after anodising and is caused by inclusions and undissolved constituents⁽³⁴⁾ (mainly rich in iron) being drawn to the surface from the periphery of the billet. To avoid this problem, good, clean billets and thorough homogenisation is required.

Anodising is a controlled electrolytic oxidation process that leads to the formation of an oxide surface coating many times thicker and more durable than the natural oxide film. A typical coating of 15-25µm is obtained using sulphuric acid electrolyte. Anodising will highlight local variations^(34,37,76) in Mg_2Si content and other inhomogeneities at the surface. Therefore if a portion of the extrudate should come into contact with a cold section of the run out table, the resulting local, rapid, precipitation of Mg_2Si will show up as patches and streaks on the extrudate surface after anodising.

Silicon precipitates on the surface cause dark spots upon anodising, though a more deleterious effect of excess silicon is in combining with iron to form AlFeSi . This phase is harmful in both brightening and anodising treatments^(35,37,75). However, the diffuse appearance due to high iron content is desired by some consumers⁽⁷⁵⁾.

Al-Mg-Si alloys are generally age hardened after extrusion and it has been shown⁽⁷⁷⁾ that the ageing treatments used also have an effect on the finishing characteristics. Naturally aged sections (T4) will be more specular than artificially aged (T6). Franz⁽⁷⁵⁾ has shown that regardless of iron content, which favours a diffuse appearance, a T6 temper always produced a more matt finish than the T52 temper, with a T4 temper producing the most specular.

If a bright finish is sought by the consumer certain chemical methods are available to achieve this. Chemical brightening is a process where a clean piece of aluminium is immersed in a solution of acids, usually a combination of chemicals including phosphoric and nitric, which dissolve the surface peaks preferentially thereby smoothing the surface. Copper has been found⁽²⁸⁾ to enhance brightness in AA6063 and AA6463 by bright dipping and anodising.

The generation of an acceptable (i.e. commercial) surface finish is therefore a complex balance of the effects of casting^(30,34), homogenisation treatments^(20,34), extrusion tooling and practices^(68,73,77,78), post extrusion handling, heat treatments^(75,77) and all chemical treatments. Obviously, the production of a sound, defect-free and bright extrudate product with

the minimum number of die lines is the most important weight in this balance.

CHAPTER THREE

CHAPTER THREE

Theory

3.1. Introduction

Analysis of extrusion through empirical and theoretical equations has been reviewed by many authors^(1-5,7-11,13,14,79,80) and therefore will not be reiterated here.

Before any extensive investigations can be made of the behaviour of a metal during extrusion, it is necessary to determine its flow stress characteristics at high temperatures.

This is most conveniently achieved using a simple testing method such as tension, compression or torsion. The tensile test has a somewhat limited applicability for modelling the extrusion process as the necking phenomenon limits the strain attainable to a typical value of 0.3. The compression test has a similar limitation imposed by the onset of barreling, but when performed on a cam-plastometer, with effective platen lubricants, a wide range of data can be obtained to strains of the order of 0.7, which is of some use in low strain, hot working processes.

The torsion test is by far the most suitable⁽⁹⁾ method to obtain, hot working, flow stress data to model the extrusion process. In ductile alloys such as AA6063, strains in excess of 20 are easily

achieved.

The theoretical stress condition of pure shear during hot torsion is complicated due to the lengthening⁽⁸¹⁾ of the specimen during testing. The change in dimensions can not be accommodated easily in the analysis, and so the specimen is restrained axially during the test. The restraint introduces an axial stress, but is preferred to a specimen geometry change. At high homologous temperatures the axial stress developed is negligible and is often ignored⁽⁵¹⁾. However, it is interesting to note that the development of the axial stress can be related⁽⁸²⁾ to the texture development during testing. As the deformation temperature is decreased, the contribution of the axial stress to the (von Mises) effective stress becomes more important and the rotation of the principal stress axes more prominent. In the present work, the axial load was measured, together with the torque and angular displacement, via a computer, and was used in all conversions into effective flow stress data. However, even at the lowest temperature (350°C), the contribution to the effective stress value was under 0.25%.

Conversion of the torque/twist (/axial load) results in to effective stress/strain data can be achieved via a number of techniques^(6,11,83-85). Two of these methods were used in the current research and are described in the following sections.

Equation (3.6.) can be further expanded using the identity:

$$\sinh^{-1}(x) = \ln(x + (1+x^2)) \quad -(3.7.)$$

The surface strain-rate ($\dot{\epsilon}$) can be expressed in terms of the twist rate, and specimen geometry by the equation:-

$$\dot{\epsilon} = \frac{2\pi \cdot r \cdot \dot{\theta}}{\sqrt{3} \cdot l} \quad -(3.8.)$$

where:

$$\begin{aligned} \dot{\theta} &= \text{Twist rate (r.p.s.)} \\ l &= \text{Specimen length (m)} \end{aligned}$$

Expanding the 'Z' term in Equation (3.6.) and substituting Equation (3.8.) allows Equation (3.4.) to be written:-

$$M = \frac{2\pi}{\sqrt{3} \cdot \alpha} \int_0^R (\ln(\Phi r^{1/n}) + \ln(1 + (\Phi r^{1/n})^{-2})^{0.5}) r^2 dr \quad -(3.9)$$

$$\text{where:} \quad \Phi = \left(\frac{2\pi \cdot \dot{\theta} \cdot \exp(\Delta H/RT)}{\sqrt{3} \cdot l \cdot A} \right)^{1/n} \quad -(3.10.)$$

Equations (3.9.) and (3.10.) now relate the torque, twist rate and temperature through four hot working constants ' ΔH ', ' α ', ' n ', and ' A '.

Equation (3.9.) can also be written as the sum of three integrals:-

$$M = \frac{2\pi}{\sqrt{3} \cdot \alpha} (J_1 + J_2 + J_3) \quad -(3.11.)$$

where:

$$\left. \begin{aligned} J_1 &= \ln \Phi \int_0^R r^2 \cdot dr \\ J_2 &= \frac{1}{n} \int_0^R r^2 \cdot \ln(r) \cdot dr \\ J_3 &= \int_0^R r^2 \cdot \ln(1 + (1 + (\Phi \cdot r^{1/n})^{-2})) \cdot dr \end{aligned} \right\} \quad -(3.12.)$$

A solution for integrals J_1 and J_2 can be found analytically; a solution to J_3 can be found in references (7,9). A computer program was written^(7,9,11) to find a solution for Equation (3.9.) using the above method from a matrix of torque, twist rate and temperature data. The temperature rise during deformation was calculated using Wright's⁽⁶⁾ finite difference model.

From a matrix of five strain-rates and five temperatures, the computer program iterated the values of ' ΔH ', ' α ', ' n ' and ' A ' to obtain a least square regression between the calculated and experimental values. Once a minimised solution was reached, the values for the hot working constants could be used to calculate the flow stress at any strain-rate and temperature using Equation (3.5.).

The conversion method described above has been used successfully for numerous aluminium alloys^(7-9,11,13,14), steel⁽¹²⁾ and titanium⁽¹⁵⁾.

Attempts to establish a strain dependence to Equation (3.5.) was made by Veirod⁽¹¹⁾ by minimising the developed torque at homologous strains between 0.5 and 1. The method however, required considerable computer time, and more importantly did not present the results in a convenient form. The increasing requirements for a strain-based constitutive equation, for use in finite element programs, necessitated an alternative conversion technique.

3.2.2. Effective Radius

As mentioned previously, the fundamental problem with the interpretation of the torque/twist data arises from the strain and strain-rate gradient across the specimen section. Fields and Backofen⁽⁸³⁾ showed that the surface shear stress could be determined from the torque measured in hot torsion tests when the dependence of torque on twist rate and angle of twist was known. An alternative approach that leads to the same functional relationship is to assume that the shear stress (τ) at any radius (r) is related to the shear strain (Ω) and shear strain-rate ($\dot{\Omega}$), at that radius by⁽⁵²⁾:-

$$\tau = K \cdot \dot{\Omega}^n \cdot \Omega^m \quad \text{---(3.13.)}$$

where: K, n, m , are constants and:-

$$\Omega = \frac{r \cdot \Theta}{l} \quad \text{---(3.14.)}$$

$$\dot{\Omega} = \frac{r \cdot \dot{\Theta}}{l} \quad \text{---(3.15.)}$$

where:
 Θ = Twist (rad)
 $\dot{\Theta}$ = Twist rate (rad.s⁻¹)
 l = Specimen length (m)

The torque can now be evaluated from:-

$$M = 2\pi \int_{r1}^{r2} r^2 \cdot \tau \cdot dr \quad \text{---(3.16.)}$$

substituting Equations (3.13.-3.15.) gives:-

$$M = 2\pi \int_{r_1}^{r_2} r^2 \cdot K \cdot \left(\frac{r \cdot \dot{\theta}}{1} \right)^n \left(\frac{r \cdot \theta}{1} \right)^m \cdot dr \quad -(3.17.)$$

$$= \frac{2\pi K (\dot{\theta}/1)^n (\theta/1)^m}{(3+m+n)} \cdot (r_2^{(3+m+n)} - r_1^{(3+m+n)}) \quad -(3.18.)$$

$$= \frac{2\pi}{(3+m+n)} \cdot (r_2^3 \tau_2 - r_1^3 \tau_1) \quad -(3.19.)$$

where τ_1 and τ_2 are the shear stresses at r_1 and r_2 respectively.

Rewriting Equation (3.18.) in terms of shear stress τ_x at a given radius r_x we get:-

$$M = \frac{2\pi \cdot \tau_x}{(3+m+n)} \cdot \frac{(r_2^{(3+m+n)} - r_1^{(3+m+n)})}{r_x^{(m+n)}} \quad -(3.20.)$$

For a solid specimen $r_1=0$ and putting $r_x=r_2$ gives the relationship between τ_{surface} and torque derived by Fields and Backofen.

Equation (3.20.) suffers the disadvantage that 'm' and 'n' must be known for the shear stress to be determined from the measured torque. Furthermore, 'n' varies with strain and 'm' with strain-rate thereby making conversion tedious even when these values are known.

To overcome this, an effective radius can be defined. If a constant shear stress (τ^0) exists throughout the section, then $m = n = 0$ and



Equation (3.19.) becomes:-

$$M = \frac{2\pi \tau^0}{3} (r_2^3 - r_1^3) \quad -(3.21.)$$

Then using Equations (3.20.) and (3.21.) at a given torque we get:-

$$\frac{\tau^0}{\tau_x} = \frac{3}{(3+m+n)} \cdot \frac{(r_2^{(3+m+n)} - r_1^{(3+m+n)})}{(r_2^3 - r_1^3) \cdot r_x^{(m+n)}} \quad -(3.22.)$$

The effective radius (r_{ef}) is then defined as the radius at which $\tau^0/\tau_x = 1$, that is:-

$$r_{ef} = \frac{3 \cdot (r_2^{(3+m+n)} - r_1^{(3+m+n)})^{-1/(m+n)}}{(3+m+n) \cdot (r_2^3 - r_1^3)} \quad -(3.23.)$$

In this present study solid specimens were used ($r_1=0$) therefore Equations (3.22.) and (3.23.) become:-

$$\frac{\tau^0}{\tau_x} = \frac{3}{(3+m+n)} \cdot \left(\frac{r_2}{r_x} \right)^{m+n} \quad -(3.24.)$$

$$r_{ef} = \left(\frac{3}{(3+m+n)} \right)^{1/(m+n)} \cdot r_2 \quad -(3.25.)$$

Using Equation (3.24.), the ratio τ^0/τ_x for $0 < r_x < r_2$ was calculated for different $(m+n)$ values. The results are shown plotted in Figure (3.1.). In each event the locus for each $(m+n)$ value passes

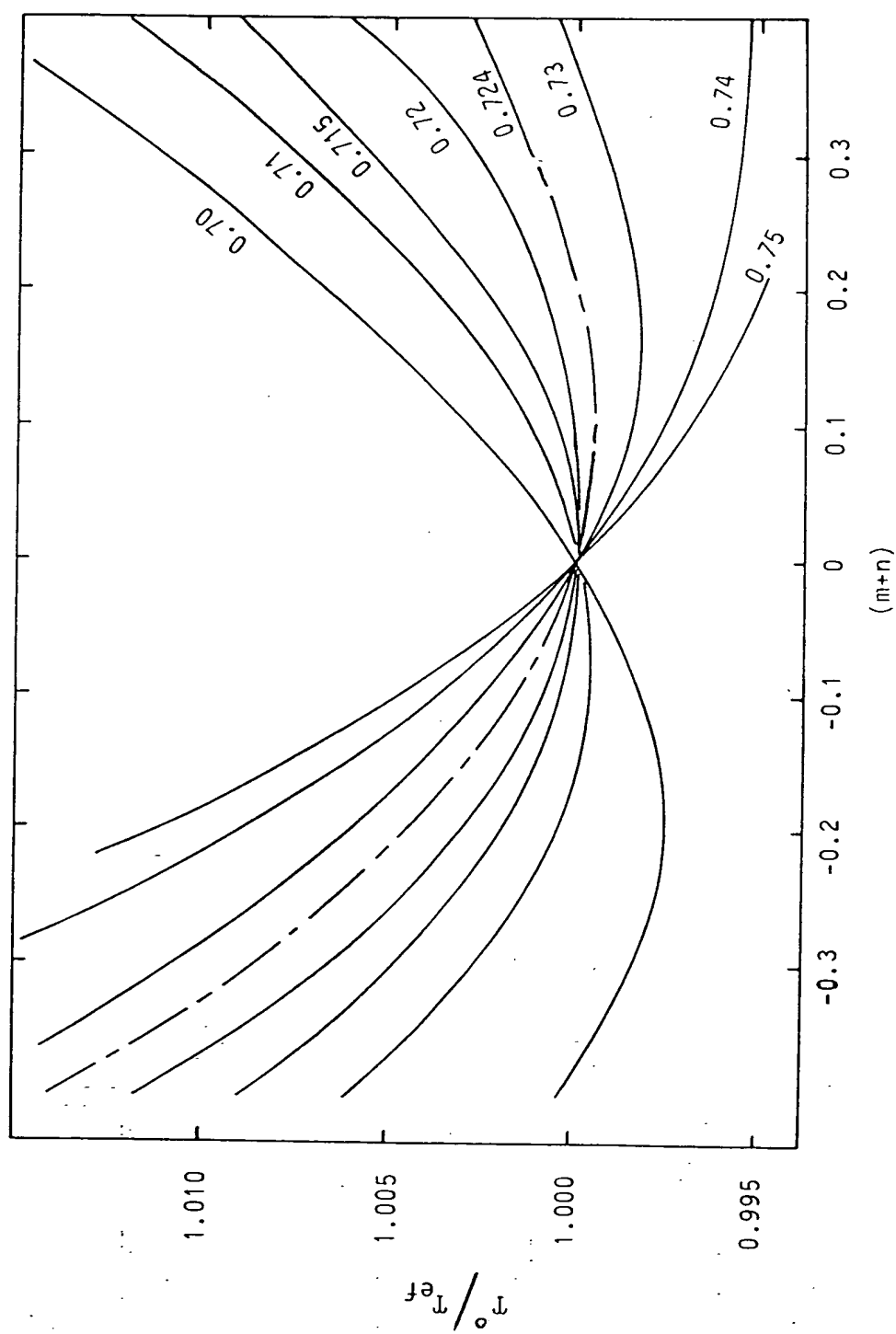


Figure 3.1. Variation of T°/T_x with $(m+n)$

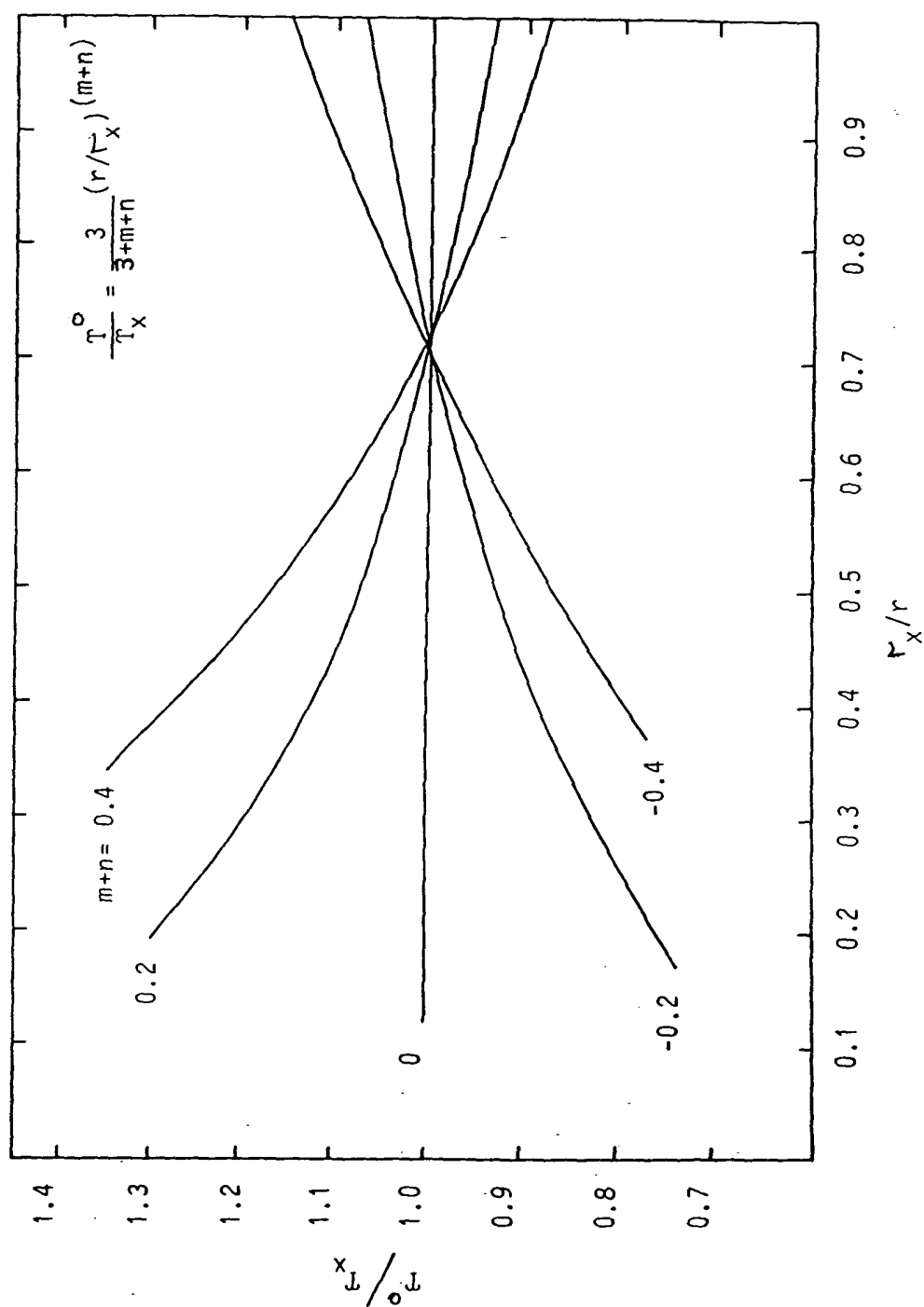


Figure 3.2. Effective Radius Position with Respect to $(m+n)$ Value

through a radius of approximately $r_{ef} = 0.72.r_2$ at $\tau^0/\tau_x = 1$. This indicates that τ^0/τ_x is almost independent of the $(m+n)$ value at the effective radius (r_{ef}). The exact value for r_{ef} can of course be calculated using Equation (3.25.). The results are shown, plotted in Figure (3.2.), for a wide range of $(m+n)$ values. It is clear that the choice of $(m+n)$ to calculate r_{ef} has little effect on its location, and in addition, causes only minor variations in the τ^0/τ_{ef} value. The value of $(m+n)$ for aluminium alloys, which do not dynamically recrystallise, is typically⁽⁸⁵⁾ in the range (0-0.2).

Because the 'effective radius' location is insensitive to the value of $(m+n)$ it is advantageous to calculate the deformation parameters, and study structural features at this location rather than at the specimen's surface.

Therefore, the deformation parameters are defined as:-

$$\begin{aligned}\tau &= \frac{3M}{2\pi.r_2^3} \\ \Omega &= \frac{r_{ef}.\Theta}{1} \\ \dot{\Omega} &= \frac{r_{ef}.\dot{\Theta}}{1}\end{aligned}\quad -(3.26.)$$

The equivalent stress and strain can be subsequently calculated using the Von Mises criterion.

The advantage of this conversion method is that the torque/twist

curve can be directly projected to a stress/strain curve. In addition, the use of a computer for data collection permitted immediate analysis of the digitised stress/strain data.

3.4. Surface Layer Temperature Analysis

The temperature rise due to billet deformation during extrusion has been estimated using computer programs, based on algorithms originally proposed by Sheppard and Raybould⁽⁸⁸⁾ and subsequently modified by other workers^(5,11,14,15). Because of the particular interest in the surface generated features in the present research, it was necessary to include the additional heat source arising from frictional condition between the die land and newly generated extrudate surface. The resultant temperature rise from this source is calculated using Stuwe's⁽⁸⁹⁾ method:-

$$T_{sr} \approx \sigma \cdot \frac{1}{4 \cdot C_p \cdot \rho} \left(\frac{s \cdot V}{a} \right)^{0.5} \quad -(3.27.)$$

where:

s = Die land length
V = Exit speed
a = Thermal diffusivity
σ = Flow stress
C_p = Specific heat
ρ = Density.

Where the heat generated penetrates a depth:-

$$Y = \left(\frac{a \cdot s}{V} \right)^{0.5} \quad -(3.28.)$$

If we were to assume that the flow stress in Equation (3.27.) can be taken as that defined by the Zenner Holloman Equation, then by

substituting into Equation (3.27.) we arrive at:-

$$T_{sr} \approx \frac{1}{4.C_p.\rho.\alpha} \cdot \sinh^{-1} \left(\frac{Z}{A} \right)^{1/n} \cdot \left(\frac{s.v}{a} \right)^{0.5} \quad -(3.29.)$$

Unfortunately, the validity of Equation (3.28.) has not been determined in the present research programme, due to difficulties in obtaining an accurate measurement of the extrudate surface temperature. However, the temperature rise estimates are consistent with observed structural and topological phenomena.

CHAPTER FOUR

CHAPTER FOUR

Experimental Procedure and Equipment

4.1. Introduction

The experimental procedures and equipment used in the course of this study are discussed in the following sections. Techniques common to other research projects (7-9,11) were developed to suit the requirements of the current research program.

4.2. The Extrusion Press

All the extrusion trials were conducted using a 5MN vertical extrusion press which was capable of operating in both the direct and indirect mode. The press was situated over a pit in which resided an air/water quench tube 1.5m in length. The main ram was propelled by a hydraulic slipper pump resulting in ram speeds of 3mms^{-1} to 14mms^{-1} during the extrusion cycle. An auxiliary pump assisted the ram movement prior to the extrusion cycle to produce a fast approach thus minimising the time for heat loss from the billet. The auxiliary pump automatically cut out once the extrusion load became greater than 0.2MN. An alternative driving system for the ram was also used in the form of an accumulator drive. This is discussed in more detail in section 4.2.2.

The container, which could be raised and lowered hydraulically, allowed the extrusion of billets with a maximum dimension of 75mm in diameter by 120mm in length. The container was heated by two sets of electrical heating elements with a combined power rating of 16kW. The maximum temperature permitted in the container was 530°C, this being governed by the tempering temperature of the steel. The container temperature, controlled via a Eurotherm, gave a temperature profile on the chamber wall of $\pm 5^{\circ}\text{C}$ of the control temperature.

A variable frequency induction heater was used for the billet preheat, yielding heating rates for AlMgSi cast billets of typically $5\text{--}10^{\circ}\text{Cs}^{-1}$. The induction heater was moved pneumatically over the container chamber once the billet had reached its test temperature, allowing the billet to be fed directly into the chamber thus minimising the load time and associated temperature losses.

The layout of the extrusion press can be seen in Plate (4.1.).

4.2.1. Accumulator Drive

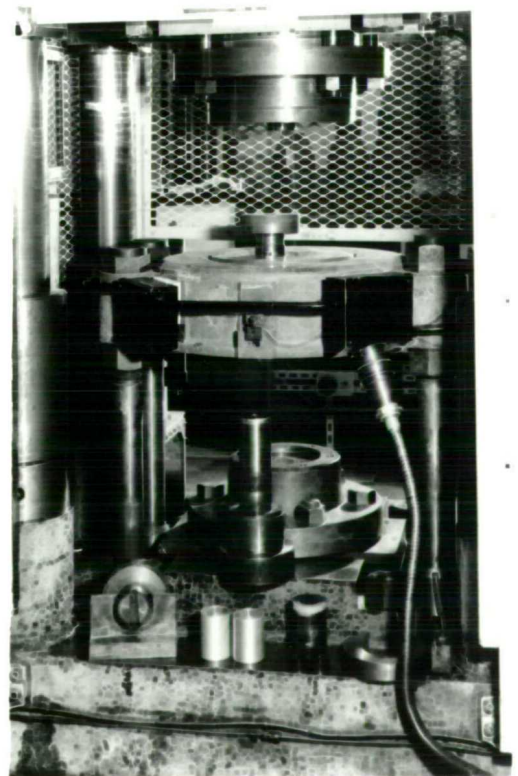
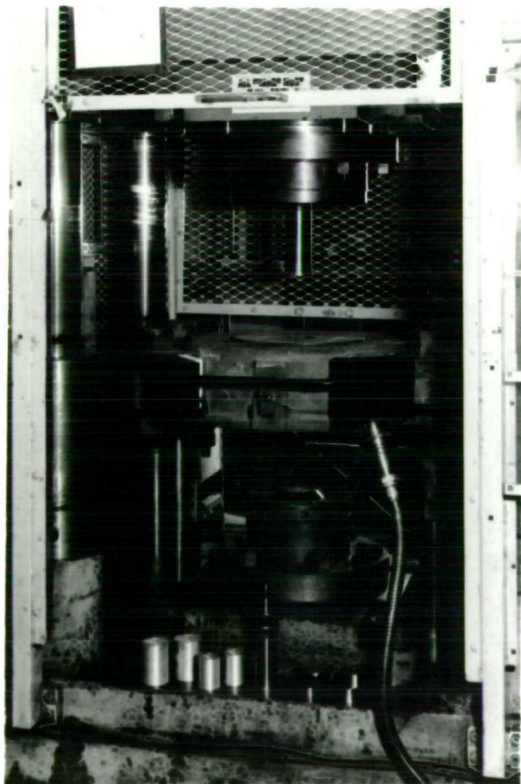
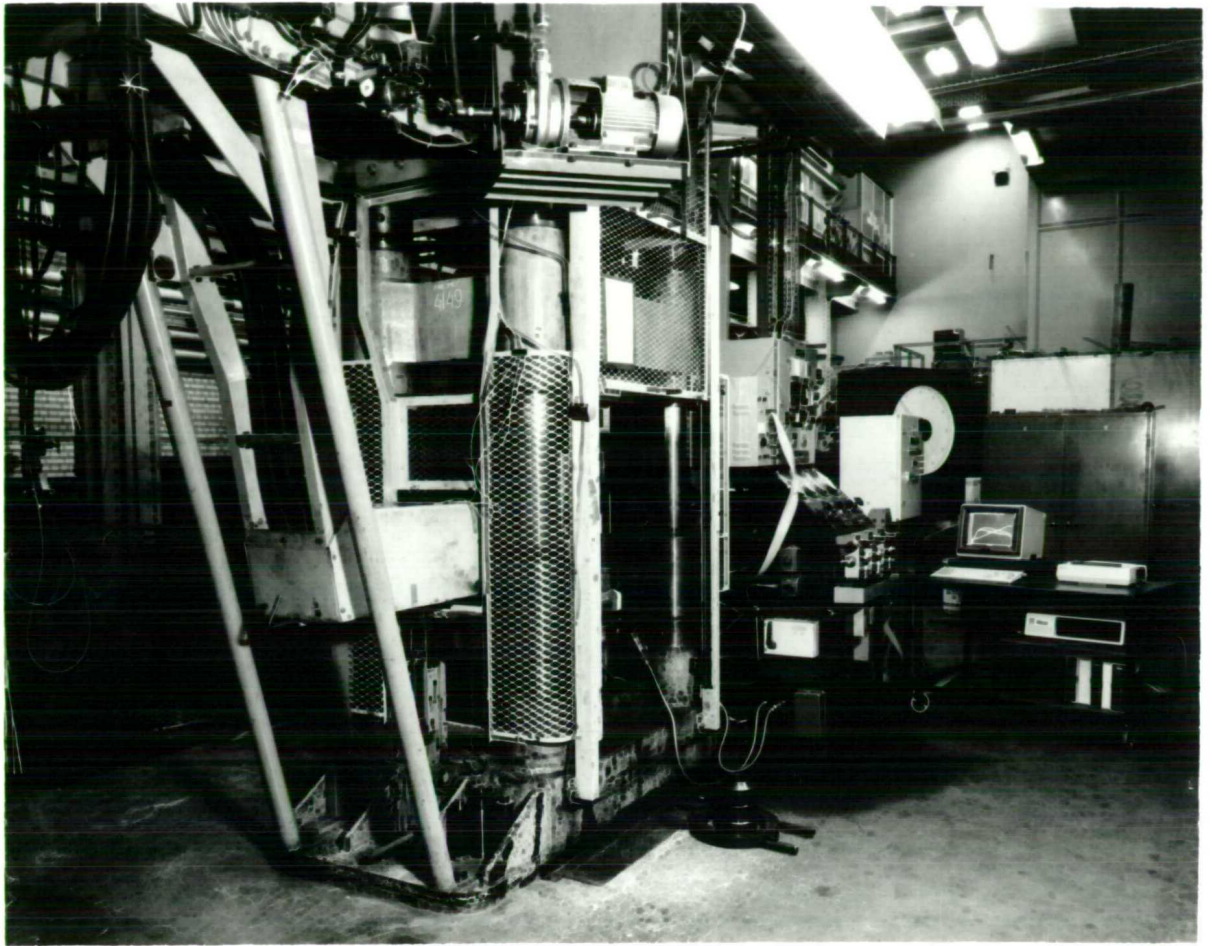
Ram speeds outside the range available from the slipper pump were achieved using an accumulator drive. The accumulator was charged prior to extrusion using four nitrogen filled bottles and the ram speed set electronically via potentiometers which selected the valve operation and arrangement. Two spool valves governed the rate of oil delivery to the main ram which could either be used individually or in combination. This resulted in four ram speed ranges and produced an overall ram speed range of 0.1mms^{-1} to 240mms^{-1} . The extrusion trial procedure when using the accumulator was similar to that for the

(a) General Layout of the Extrusion Press

(b) Direct Tooling

(c) Indirect Tooling

Plate 4.1. 5MN Extrusion Press.



slipper pump drive. An auxiliary pump enabled a fast approach by the ram which automatically cut out when the load reached 0.2MN. The ram then proceeded at a speed governed by the slipper pump until a load of 1MN was reached. At this point, the accumulator 'cut in' and the ram proceeded at a speed set prior to the trial.

4.2.2. Direct Extrusion Tooling

Plate (4.1.a) shows the direct extrusion tooling for 75mm in diameter billets. In the foreground 90mm x 75mm ϕ billets, pressure pad, scraper pad and dies can be seen, and in the background the ram, container and die assembly. Also shown are 70mm x 57mm ϕ billets and ram. The alternative 57mm ϕ extrusion tooling allows the maximum extrusion pressure to be increased by almost 75% which is sometimes necessary for some highly alloyed aluminium powder billets (e.g. Al-5Cr-2Zr). The ram, container liner, dies, pressure pads and scraper pads were all machined from KEA 145 5% Cr-V steel.

Changing the die was achieved by taking off the retaining ring and lifting out the die assembly which enabled the die backer and die to then be removed. These components are shown schematically in Figure (4.1.a). The dies used for direct extrusion trials were for rod shaped extrudates with reduction ratios of 10, 20, 30, 50, 60, 80, 100 and 150:1. In addition, a split die tooling arrangement was used to study the surface generation aspect of aluminium extrusion. The split die tooling assembly, shown schematically in Figure (4.2.), allowed the die halves to be removed after the extrusion trial had stopped without damaging the extrudate surface, permitting inspection of the die land

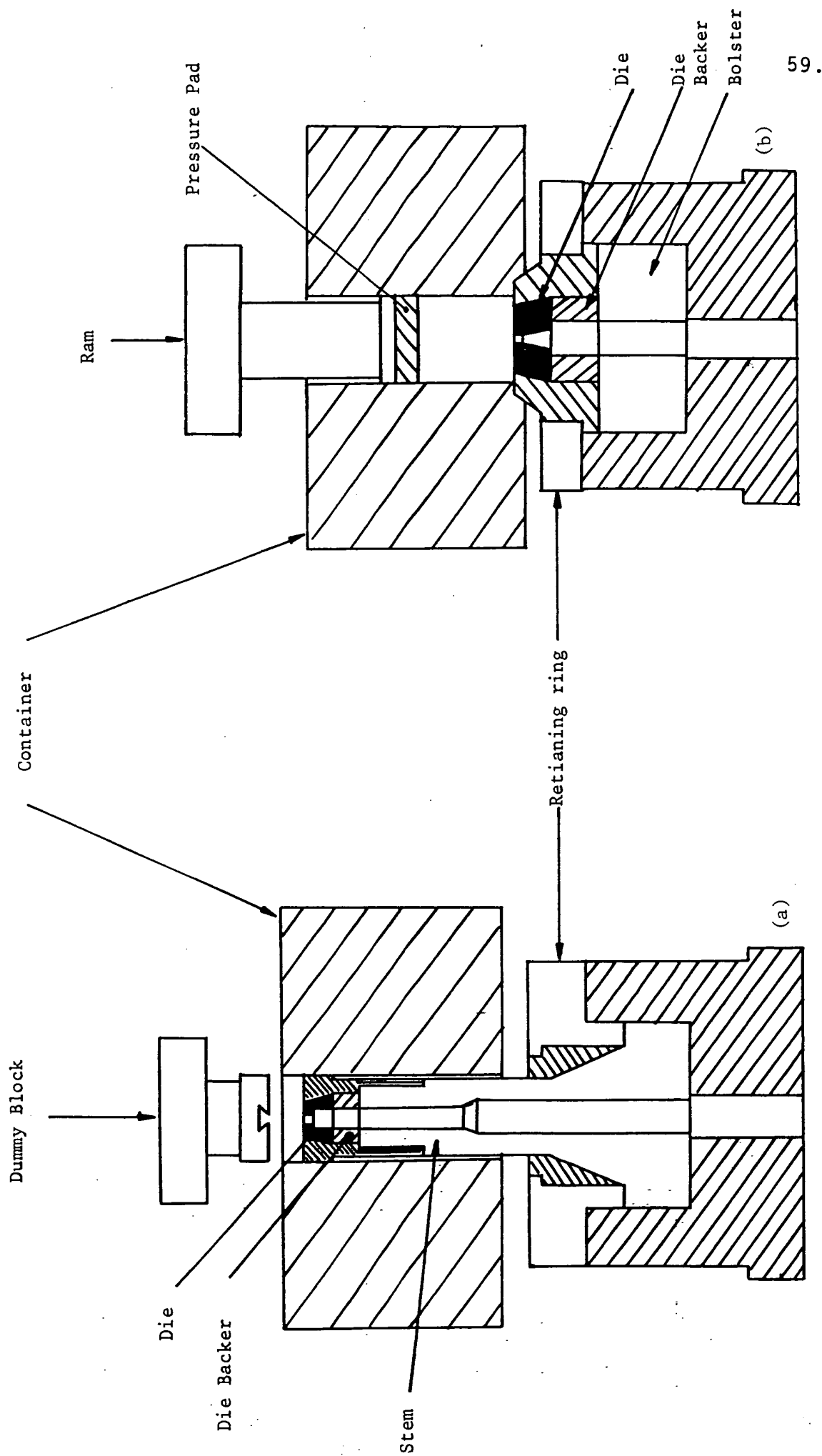


Figure 4.1. Indirect and Direct Tooling (Schematic)

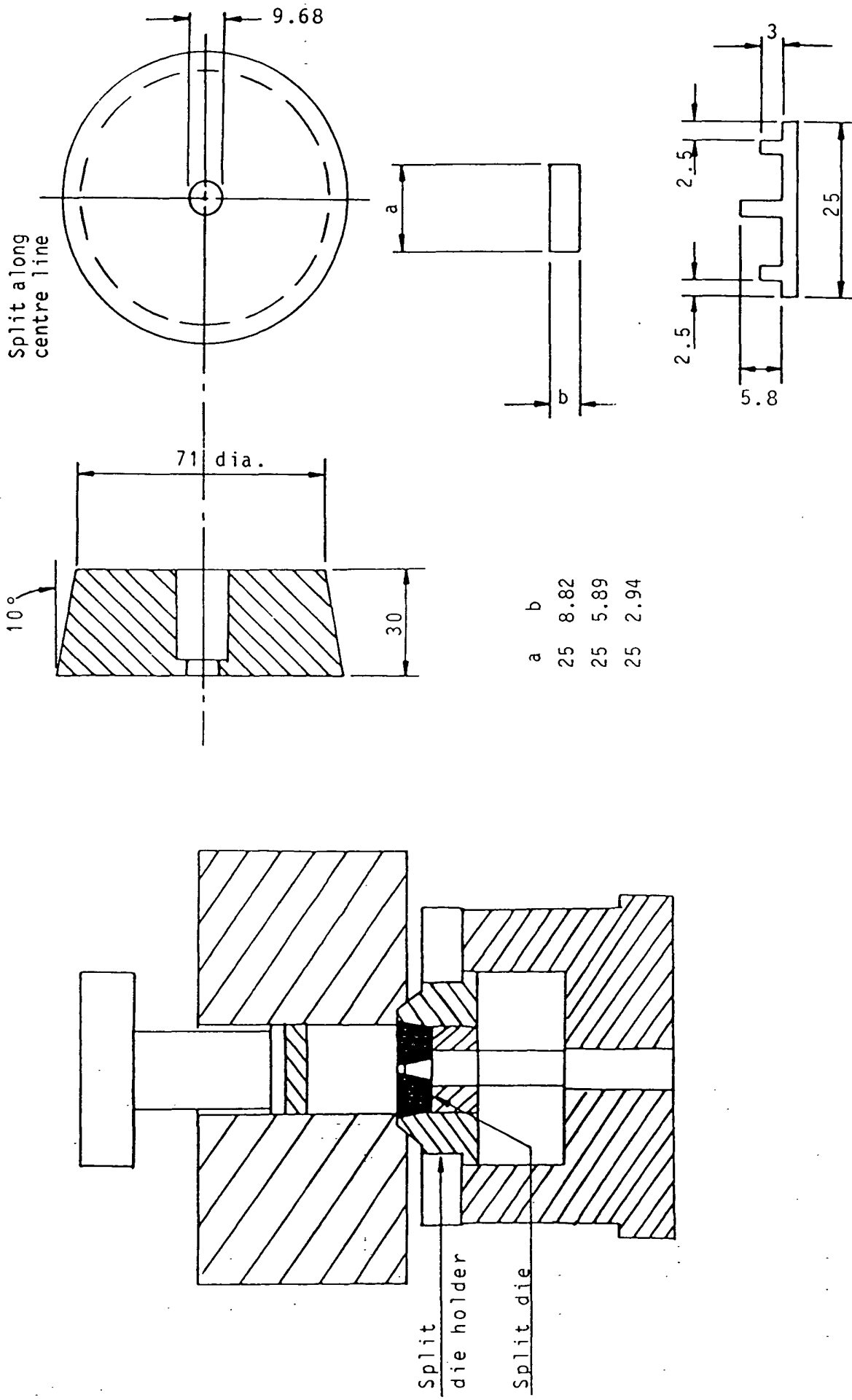


Figure 4.2. Split Die Design and Tooling

area. The dies had strip (20, 30 and 60:1) and ribbed (60:1) sections. The dimensions of these sections are given in Figure 4.2 .

To study the effect of die land length on extrudate surface generation, a series of dies with die land lengths of 5, 4, 3, 2, 1mm and knife edge were used for both direct and indirect extrusion modes.

4.2.3. Indirect Extrusion Tooling

The indirect extrusion tooling is shown in Plate (4.1.b). The ram and pressure pad have been replaced by a dummy block with a tapered notch for the removal of the discard after extrusion (shown to the left in the foreground in Plate (4.1.b)). The die assembly is located at the top of a mandrel, the dimensions of which allow the container to pass over it. The mandrel and die assembly are shown in the centre of Plate (4.1.b) and have been removed from their operating position. A spare die holder, die backer and die are also shown in the foreground with two 90mm x 75mm ϕ billets. The horseshoe-shaped component indicated to the lower right of the Plate facilitates the removal of the discard from the container. All indirect components were machined from the hot working tooling steel KEA 145 5% Cr-V.

Changing the die was effected by lifting the container, unscrewing the die holder and removing both the die backer and die. The dies used for indirect extrusion were rod shaped with reduction ratios of 20, 30, 40, and 80:1.

The dimensional arrangement of the tooling for indirect extrusion was

such that only 75mm ϕ billet extrusions were possible with a minimum reduction ratio of 20:1. The tooling arrangement for the indirect mode is shown schematically in Figure (4.1.b).

4.2.4. Extrudate Quenching

In order that the structure of the material should reflect that due to the hot working process, the extruded product was passed through a water quench immediately upon exit from the die assembly.

The quench consisted of two concentric cylinders, the inner containing holes of 1.2mm ϕ drilled radially, 532 of which were contained in the initial 0.26m, the remainder in groups of 28 holes placed at 0.1m intervals along the 1.37m length. Water was delivered at a rate of 5 litres per second by a centrifugal pump. Experiments were conducted to measure the cooling rate for various extrudate diameters which yielded a typical heat transfer coefficient of $h = 325 \text{ W/m}^2\text{°C}$.

4.2.5. Experimental Procedure

Certain components and procedures were common to direct and indirect extrusion. The container was left at the required temperature (50°C below that of the billet) for at least thirty minutes before extrusion to allow the die and bolster to attain a steady temperature. The pressure pad, or the dummy block used in the indirect mode, was also preheated on the container for 30 minutes prior to the extrusion trial. The billets were then extruded via flat faced dies without

lubrication by both extrusion modes.

4.2.6. Direct Extrusion

Prior to the extrusion trial, the container was hydraulically lowered onto the die assembly to obtain a pressure seal. Two semi-circular rings were placed on top of the container to prevent any damage to the main ram. Once the billet had been transferred from the induction heater to the container, the pressure pad was dropped onto the billet. The fast ram approach and water quench were then initiated. Immediately before the ram hit the billet a micro switch was activated which set-up the computer data recording. Extrusion was then continued at the pre-determined speed (using either slipper pump or accumulator drive) until the main ram touched the rings whereupon the pumps were reversed and the ram raised. Temperature loss during billet transfer was estimated by recording the time from cessation of billet heating to the point where the billet began to upset.

On completion of the extrusion trial, the container was raised and the extrudate cut from the discard leaving a discard stem of approximately 50mm in length. The discard was then removed from the container by pushing a scraper (or scavenger) pad through the container bore, thereby, removing excess aluminium build-up from the container wall, before forcing the discard and pressure pad into a set of rings placed between the container and die assembly.

4.2.7 Indirect Extrusion

The procedure for indirect extrusion was essentially similar to that for the direct mode except that the 75mm ϕ ram was removed from the main ram and the container position was such that the die assembly at the top of the mandrel resided in the bottom 50mm of the container. Billet transfer from the induction heater was made manually due to the container position causing an obstruction for automatic transfer. Upon transferring the preheated billet, the dummy block was placed in the container and the extrusion cycle was initiated as for direct extrusion. When the main ram hit the dummy block both the billet and container were pushed down onto the mandrel, moving simultaneously at the predetermined speed.

Once the extrusion trial had been completed, the main ram was raised and the container lowered to reveal the billet discard and mandrel (die) top. The horseshoe-shaped component (Plate (4.1.b)) was then placed between the dummy block and the container top face. The discard was then parted from the die face by raising the container which allowed the extrudate to be cut from the discard. To facilitate ease of removal of the discard from the dummy block it was coated with DAG 580 prior to extrusion.

4.2.8. Extrusion Data Recording

The data from the extrusion trials were recorded using an IBM personal computer with a PC-Mate Labmaster, consisting of a mother board which

plugs into the IBM bus and a daughter board which can reside outside the computer, and is connected to the mother board via a ribbon cable. The Labmaster enables real time, high speed analog to digital(A/D) and digital to analog (D/A) conversions. The standard board can be jumpered for 16 single ended or 8 true differential inputs. The input range on the standard board is jumper selectable between 0 to 10v or -10 to 10v. Once the daughter board has performed the (A/D) conversion, the data is latched providing the capacity to start the next conversion while writing the prior conversion to the computer memory. The standard board has a maximum conversion rate of 30KHz.

A computer program was written to record the load, pressure and ram displacement during extrusion trials. The Labmaster board was jumpered to give a differential voltage input, and an input range of 0 to 10v. The jumper selections, operation amplifier circuits, and program details are given in Appendix 1.

4.3. Examination of Extrudates

Specimens for examination under optical and electron microscopes were taken from the extrudate produced during steady state deformation. The rear 300mm and 450mm in direct and indirect modes respectively, were discarded as this portion failed to reach the water quench.

4.3.1. Optical Microscopy

Specimens were cut both longitudinally and transversely from the

extrudate and mounted in Metaserv type SW and FT resins. The specimens were then ground from 120 to 1200 grade silicon carbide paper and polished on diamond wheels from 14 μ m to 1 μ m. These were then electrolytically etched using Barkers Reagent:

46ml	Tetra Fluoroboric Acid
7g	Boric Acid
970ml	Water.

With the specimen as anode and using a stainless steel cathode, a potential of 20v was applied with a maximum current density of 0.2 amps.cm⁻². The potential was applied for a minimum of 120 seconds after which the specimen was rapidly removed and washed in methanol. The specimens were then viewed under crossed polarisers to reveal the grain structure.

To examine precipitate distribution the specimens were prepared in a similar fashion to the above and etched for 5 to 30 seconds in 0.5% Hydrofluoric acid in water.

4.3.2. Electron Microscopy

Transverse and longitudinal slices of about 500 μ m thickness were carefully taken from the extrudate using an Isomet low speed wafering wheel. These foils were subsequently ground on 1200 grade silicon carbide paper to 300 μ m. Discs of 3mms in diameter were then punched from the foil and electropolished in a Struers jet thinner using a solution of 30% nitric acid in methanol at a potential of 15v and a

temperature of -30°C . The specimens were then examined in a high resolution Philips EM 301 (100kv) transmission electron microscope, (TEM), and in a Jeol Temscan (100kv) with quantitative EDX analysis.

Sample preparation of surface specimens, for die line and micro-die line study, was conducted in a similar fashion described above. However, so that the surface features were retained, the sample was only ground from the bulk extrudate side, and during jet thinning an acetate washer was placed between the specimen holder and the surface side of the sample foil to prevent electropolishing from this side. The use of an acetate washer permitted the jet thinners detector to still function.

Scanning electron microscopy (SEM) was used to examine both structural detail and topographical features of the extrudate surface. Structural examinations were made in a Joel T200 (25kv) SEM using a backscattered electron imaging (BEI) detector. Samples for the SEM structural study were polished to $0.25\mu\text{m}$ (or electropolished) and examined unetched. For topographical features the SEM was used in the secondary electron imaging (SEI) mode. Bulk samples of the extrudate surface were removed and cleaned in an ultrasonic cleaner immersed in acetone. To prevent the samples from electrically charging while under examination, they were sputtered with a fine coating of gold.

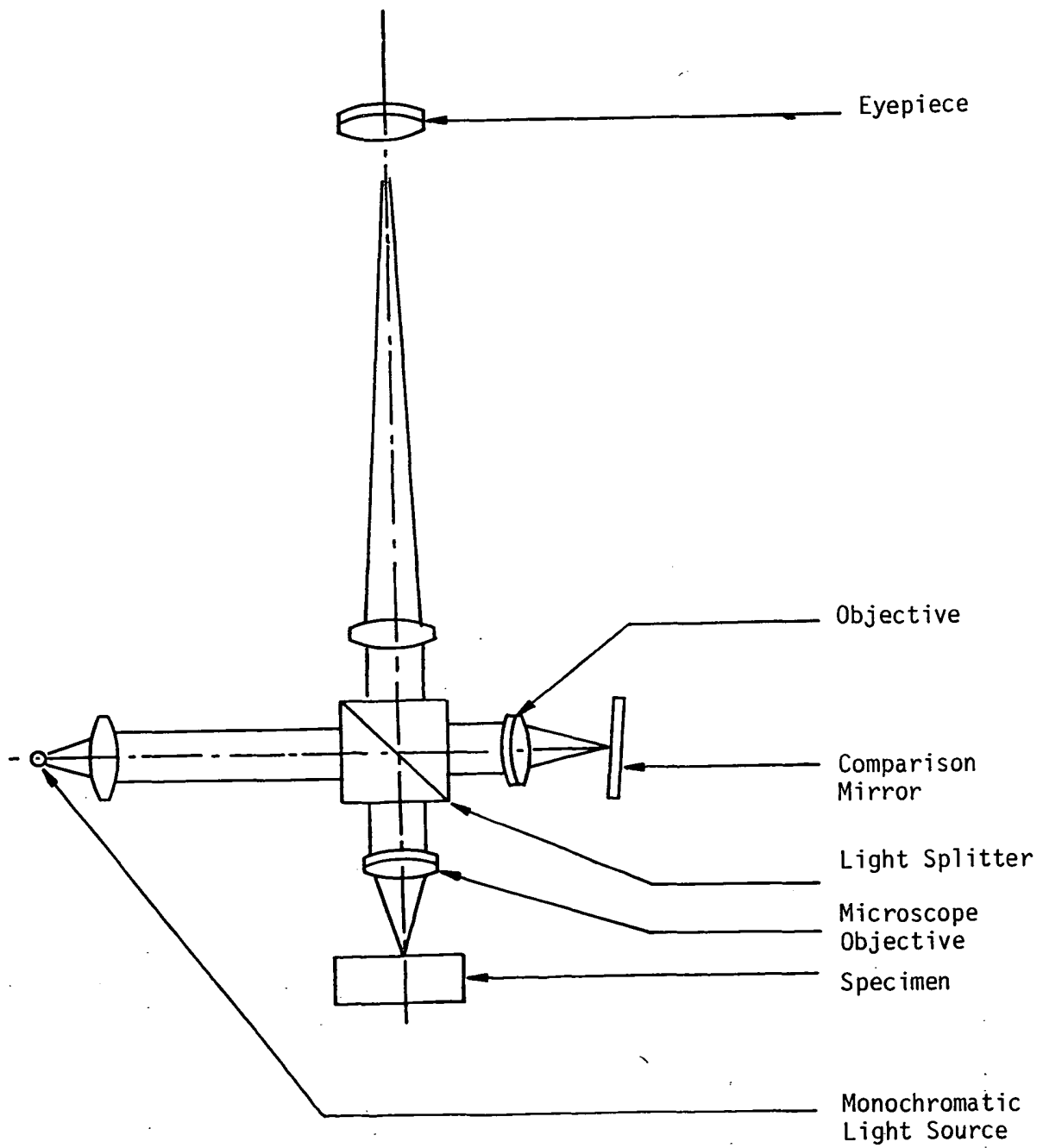
In both imaging modes an EDX qualitative analysis probe was available to assist identification of second phases and pick-up particles.

4.4. Surface Quality and Finish

Good surface finish is a difficult parameter to define as it can refer to so many different aspects. Roughness and brightness are the most commonly used parameters. However, their uses are limited as surface defects after anodising and random surface flaws may not be detected in the roughness or brightness analysis. Visual examination of the final product is often the only means of quality control. In the present work, emphasis was placed on the as-extruded surface roughness and the occurrence of pick-up. The surface roughness was assessed using two techniques. The first was the traditional method of evaluating surface roughness with a stylus type instrument (Talysurf 10). The vertical displacements of the stylus sensor as it traverses the surface are transformed into voltage variations and then amplified and recorded. The Talysurf equipment automatically integrates the results to give an 'Ra' value which corresponds to the arithmetic mean of the departures of the profiles from the mean line. The main disadvantage of this type of apparatus is the contact between the point of the stylus and the surface, which may cause damage, especially in soft aluminium alloys such as AA6063. This problem was overcome in the present work by using Acrulite replica resin.

The second technique to measure surface roughness was a polarisation interferometric method using an incident light microscope. This method of surface examination has the advantage of no surface contact and hence damage. A Reichert optical microscope was converted to an interference microscope and is shown schematically in Figure (4.3.). The interference fringes produced by the topography of the surface were

Figure 4.3. Schematic of Interference Microscope.



photographed and subsequently used to calculate the magnitude of the surface roughness by using the equation:

$$d = \frac{\lambda}{2n}$$

where: d = depth of disturbance in the specimen surface
 λ = wave length of the monochromatic light source (nm)
 n = inverse of the interference band displacement

The specimens for examination were cleaned in acetone in an ultrasonic cleaner and were examined with monochromatic light by inserting interference filters centred at 649nm thereby exhibiting high contrast interference bands.

4.5. Partial Extrusions

To examine the flow patterns during steady state direct and indirect extrusion, billets were partially extruded to a position of quasi-static deformation. The partially extruded discard was then rapidly removed from the container and water quenched. The billets were then sectioned along the longitudinal axis using a Metaserv cut-off wheel, which was continuously lubricated to avoid excessive heat generation. The face of one half of the billet was ground and polished in a similar manner to that of the extrudates and etched in Poulton reagent to reveal the grain structure and flow lines. The remaining half had a slice 500 μ m thick removed parallel to the section face using an Iso-met, low speed wafering wheel. Disc specimens of 3mm in diameter were then punched from locations of interest and prepared for TEM examination.

Partial extrusions were also performed using gridded billets. In this event, billets were sectioned along a longitudinal axis prior to extrusion and a square grid machined on one half. The halves were then coated with DAG 580 (to prevent both halves from welding during extrusion), rematched and spot welded. The prepared billet was then partially extruded and rapidly removed from the container and water quenched. The halves were subsequently prised apart to reveal the distorted grid pattern.

4.6. Extrudate Heat Treatment

AA6063 is a heat treatable alloy capable of reaching full T6 properties. In order to study the effects of heat treatments on extrudate tensile properties, various heat treatments were performed.

The classification of the various heat treatments together with the time/temperature conditions are given below:-

T1	- As extruded, quenched and naturally aged (no treatment)
T4	- Solution treated, quenched and naturally aged (525°C for 1 hour and water quenched)
T6	- Solution treated, quenched and artificially aged (525°C for 1 hour and water quenched) (Then; 180°C for 8hrs)
0	- Anneal (410°C for 3 hours and cooled at 10°C/hr to 260°C).

4.7. Tensile Testing

Standard No.12 Hounsfield specimens were machined from extrudates in the as-extruded, naturally aged and artificially aged conditions. Tests were conducted on an Instron universal testing machine at a constant crosshead speed of 0.5mm.min^{-1} .

4.8. Materials

All materials used in the current research programme were semi-continuous (direct chill) cast logs, 75mm in diameter and were supplied by British Alcan, Banbury. The compositional analysis of each alloy is given in Table (4.1). The main alloy under consideration had compositional limits within the Aluminium Association's specification for AA6063. A small amount of commercial purity and super purity aluminium was also used in the surface generation trials.

The AA6063 was homogenised using a short industrial treatment of $575^{\circ}\text{C} \pm 5\%$ for 6 hours and cooled at 150°C/hr s to 250°C . The logs were then left to cool to room temperature, and subsequently machined to 73.5mm in diameter, cut to the required length, and faced.

4.9. Torsion Testing

Specimens for torsion testing were machined from homogenised billets to the specifications indicated in Figure (4.4.)

Table 4.1. Material Specification

Ident.	Alloy	Cu	Mg	Si	Mn	Fe	Zn	Ti	B
106D-C1	AA6063	.001	.45	.42	.07	.21	.02	.011	.001
106D-C2	AA6063	.001	.45	.42	.07	.21	.02	.010	.001
157D	AA6063	.003	.48	.42	.07	.17	.02	.009	--
381D	AA6063	.004	.53	.45	.07	.20	.02	.009	.0004
381DA	AA6063	.004	.53	.45	.07	.20	.02	.009	.0004
381DB	AA6063	.004	.53	.45	.07	.20	.02	.009	.0004
381DC	AA6063	.004	.53	.45	.07	.20	.02	.009	.0004
379D	SP A1	.003	.001	.001	--	.001	--	.0001	--
379DA	SP A1	.003	.001	.001	--	.001	--	.0001	--
380D	SP A1	.003	.002	.001	--	.001	--	.0026	.0003
380DA	SP A1	.003	.002	.001	--	.001	--	.0026	.0003

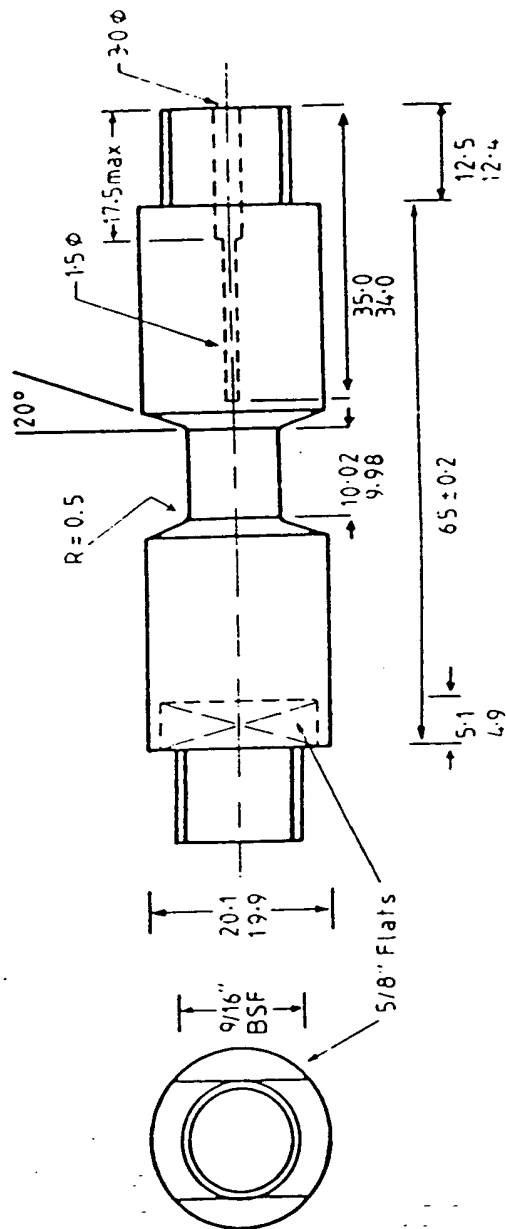


Figure 4.4. Torsion Specimen Specification

The testing machine was capable of twisting at 1000rpm maximum which resulted in a maximum surface shear strain-rate of 50s^{-1} for the specimen geometry used. Tests were performed over a range of shear strain-rates from 0.05s^{-1} to 50s^{-1} and over a temperature range of 350°C to 550°C .

The specimens were heated in-situ by an induction coil, the temperature being controlled by a mineral insulated NiCr/NiAl thermocouple inserted in one end of the specimens and connected to a Eurotherm controller. The heating rate for all the tests was set at $100^{\circ}\text{C. min}^{-1}$.

The torque generated was detected by a load cell situated at the end of a lever arm and the rotational displacement measured by a transducer located on the drive shaft of the testing machine. A load cell was also positioned between the specimen clamp and the end plate to detect any axial load development during the test. The outputs from the load cells and transducer were then amplified using standard operational amplifier circuits and subsequently recorded using a program written for the IBM PC with a PC-Mate Labmaster. The Labmaster was configured in a similar manner to that for extrusion data recording apart from the switch settings on the daughter board. These are given in Appendix 1.

Once the test was completed, the torsion specimen was automatically water quenched at a rate of $120^{\circ}\text{Cs}^{-1}$.

CHAPTER FIVE

CHAPTER FIVE

Extrusion Analysis and Constitutive Modelling

Using Hot Torsion Testing

5.1. Introduction

The deformation behaviour of aluminium during hot extrusion is complex and presents considerable difficulty in the interpretation of the results. This applies to both strength and structural aspects. So that a suitable understanding might be obtained, numerous extrusions must be performed over the range of temperature, strain-rate and reduction ratios of interest.

This process is both time-consuming and expensive, and is often impractical for alloy development programmes and alloy batch assessment.

A solution to the problem might be to use a simple testing method to model the extrusion process. In section 3.1. three such methods were discussed, and it was concluded that the torsion test was most suitable for this purpose. In the following sections, test data from the hot torsion test are analysed and presented in the form of constitutive equations. The Zenner-Holloman equation is subsequently used in the analysis of the extrusion process

5.2. Torsion Testing

The torsion test is of great utility for determining the flow stress at high strains. It is the ability to impart high accumulated strains in excess of 5 at elevated temperatures which allows the application of results to model the extrusion process. The test can also be used to assess a material's hot workability by measuring the strain to failure. In the present work, the hot torsion test has been used to evaluate the hot working characteristics of AA6063. The temperature range examined was 350°C to 550°C and strain-rates of 0.025s^{-1} to 50s^{-1} . The methods used to convert raw torque/twist data into effective stress/strain data are described in Chapter 3 and the instrumentation and experimental procedure in Chapter 4.

5.2.1. Torque/Twist Curves

The torque/twist curves from the matrix of test conditions produced two characteristic sets of curves and are shown schematically in Figure (5.1.). The first type of curve is produced at the lower strain-rate condition ($<5\text{s}^{-1}$) and at all temperatures. The torque rises to a maximum, moving through a transition of decreasing work hardening until steady state is reached where the work hardening is balanced by dynamic recovery. (Curve 'A' in Figure (5.1.)) The interval of work hardening becomes more prominent with reduction in temperature. The second type of curve ('B' in Figure (5.1.)) occurred at low temperature and high strain-rate (high 'Z'). In this event, the

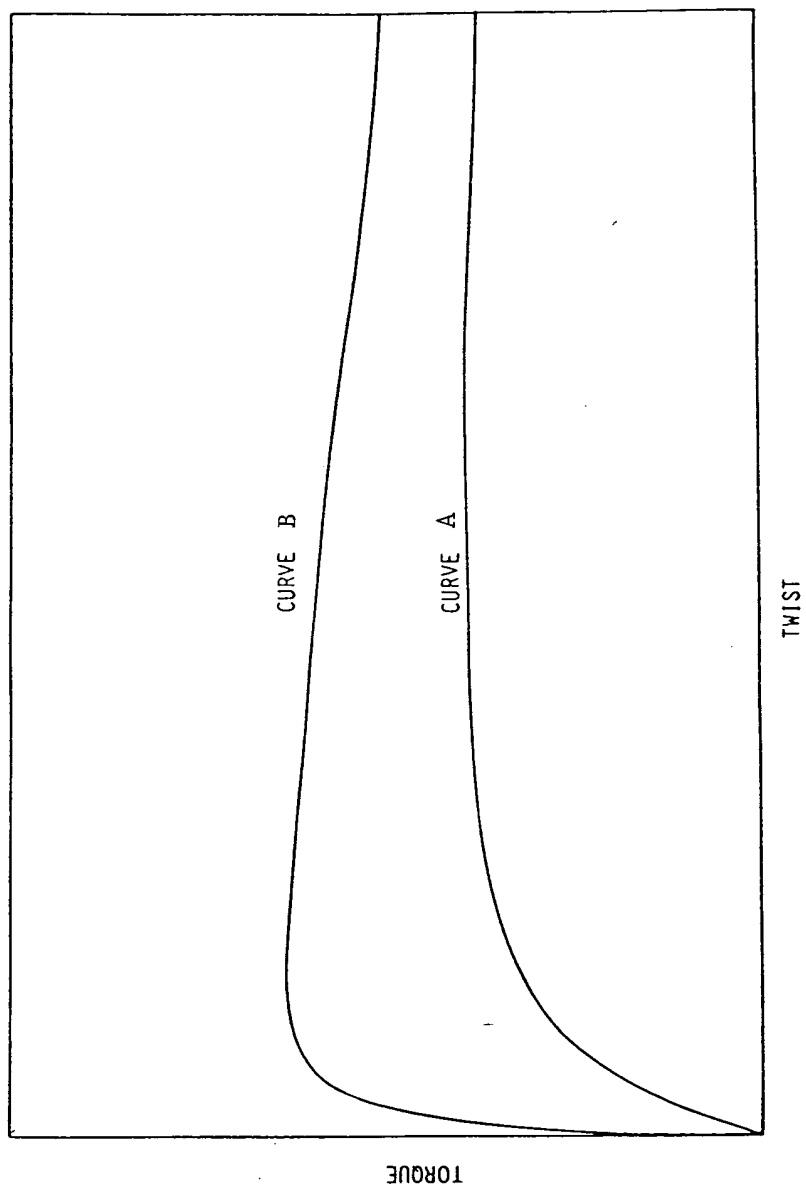


Figure 5.1 Schematic Torque Vs Twist Curves

torque, after rising to a maximum, falls steadily before failure. The drop in the torque can be attributed to the near adiabatic heating due to high strain-rate deformation.

The data from the hot torsion trials are listed in Table (5.1.). It is evident from these results that both temperature and strain-rate have a pronounced effect on the peak torque. (Demonstrated in Figure (5.2.)) This is because the torque is an integral function of the flow stress.

5.2.2. Torsion Data Analysis

The difficulty in determining the shear stress in hot torsion arises from the problem of extracting it from the integral:-

$$M = 2\pi \int_0^R \tau \cdot r^2 \cdot dr \quad \text{---(5.1.)}$$

where:	M = Torque	(Nm)
	R = Outer radius	(m)
	r = Local radius	(m)
	τ = Shear stress	(Nm ⁻²)

This is because, unlike at room temperature where ' τ ' may be considered constant across the section, at high homologous temperatures ' τ ' varies from zero at the centre, to a maximum at the specimen surface. However, numerous methods are available^(6,11,83-85) to solve Equation (5.1.). In the following sections, two of these conversion methods have been applied to the torsion data for AA6063. These techniques are fully described in Chapter 3.

It is also necessary to estimate the temperature rise due to

Table 5.1.

Torque (Nm)	Twist Rate (rad. ⁻¹)	Initial Temperature (°C)	Temperature Rise (°C)
6.8950	1.00	350	-
5.3821	1.05	400	-
3.7876	1.00	450	-
2.1175	1.10	500	-
3.6183	0.20	400	-
6.4008	0.22	350	-
5.3806	9.14	400	2.0
7.9122	9.43	350	3.5
4.2365	9.09	450	1.0
3.2314	9.52	500	0.5
8.0407	31.29	350	6.5
7.4739	93.30	350	11.5
5.9701	93.30	400	9.0
6.2875	30.10	400	4.5
4.7564	30.62	450	2.5
4.7987	93.78	450	6.5
3.6878	30.57	500	1.5
4.0355	92.34	550	4.0
3.1664	30.14	550	0.5
2.5890	9.86	550	-
1.9663	0.70	550	-

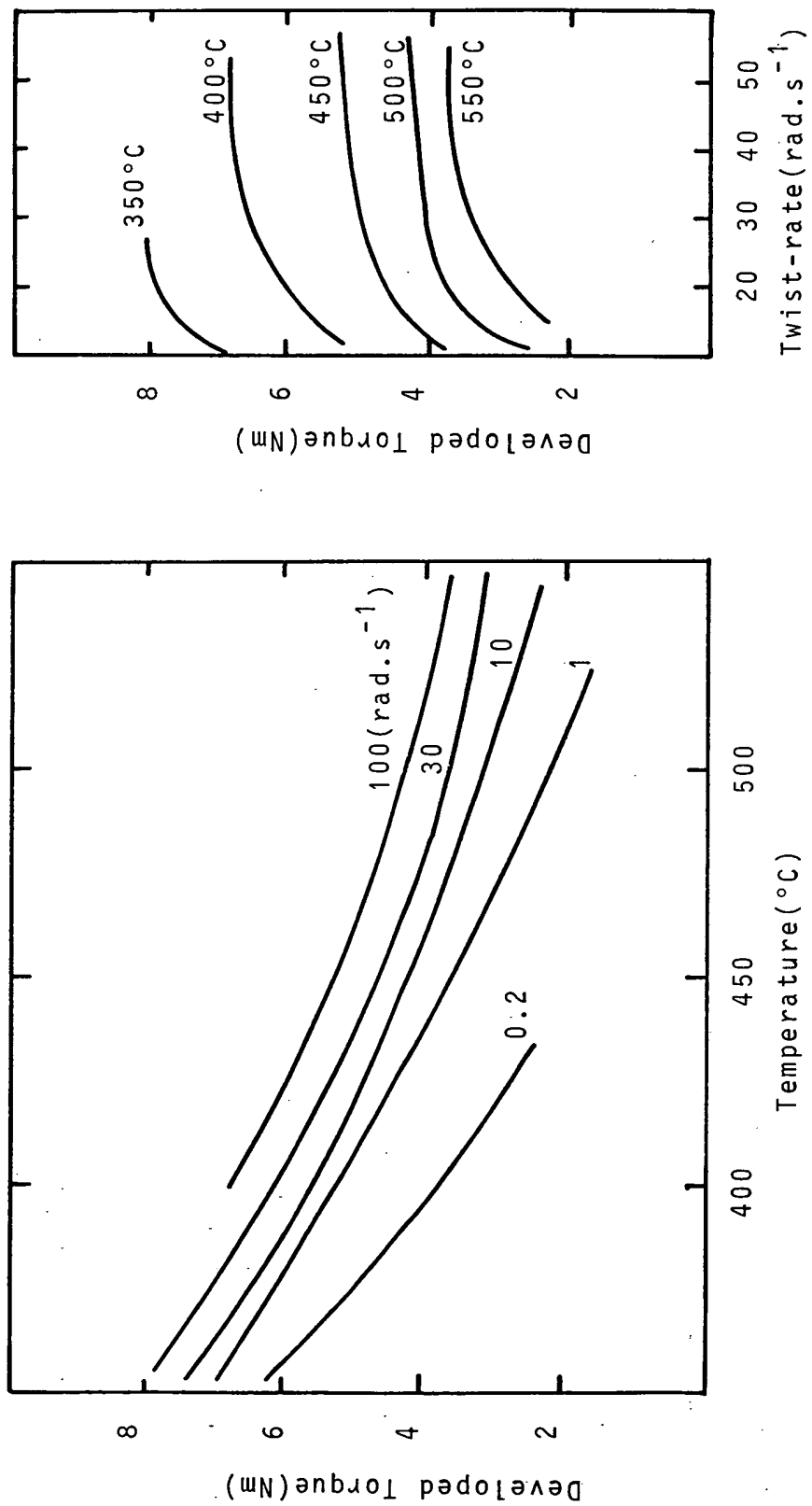


Figure 5.2. Torque Dependence on Temperature and Strain-Rate

deformation in order that an accurate solution might be obtained. This was achieved using Wright's⁽⁶⁾ finite difference model. The predicted temperature rises from this analysis are shown in Figure (5.3.). It can be seen that the rise in temperature to peak torque is small and even at the highest 'Z' condition, represents a temperature increase of under 4%. However, it was found that inclusion of the temperature rise in the torsion analysis improved the correlation when using the first conversion method (i.e. minimisation of the general hot working equation).

The small improvement in the solution for the present alloy does not emphasise the importance of including the temperature rise. However, marked improvements in solution have been reported^(11,14) in high strength aluminium alloys where the temperature increase becomes more significant.

5.2.2.1. Torque Minimisation

The results from the conversion of torsion data into flow stress data using minimisation of the general hot working equation are listed in Table (5.2.) and are plotted in Figure (5.4.) at an homologous strain of ' $\epsilon_h = 1$ '. The results demonstrate the fundamental features of the flow stress at hot working temperatures. At a constant strain-rate there is a decrease in the flow stress with increase in temperature. This trend is repeated at each constant strain-rate but is transposed to a higher magnitude of stress with increasing constant strain-rate. A similar feature is observed at constant temperature. In this event, the flow stress decreases with decrease in strain-rate.

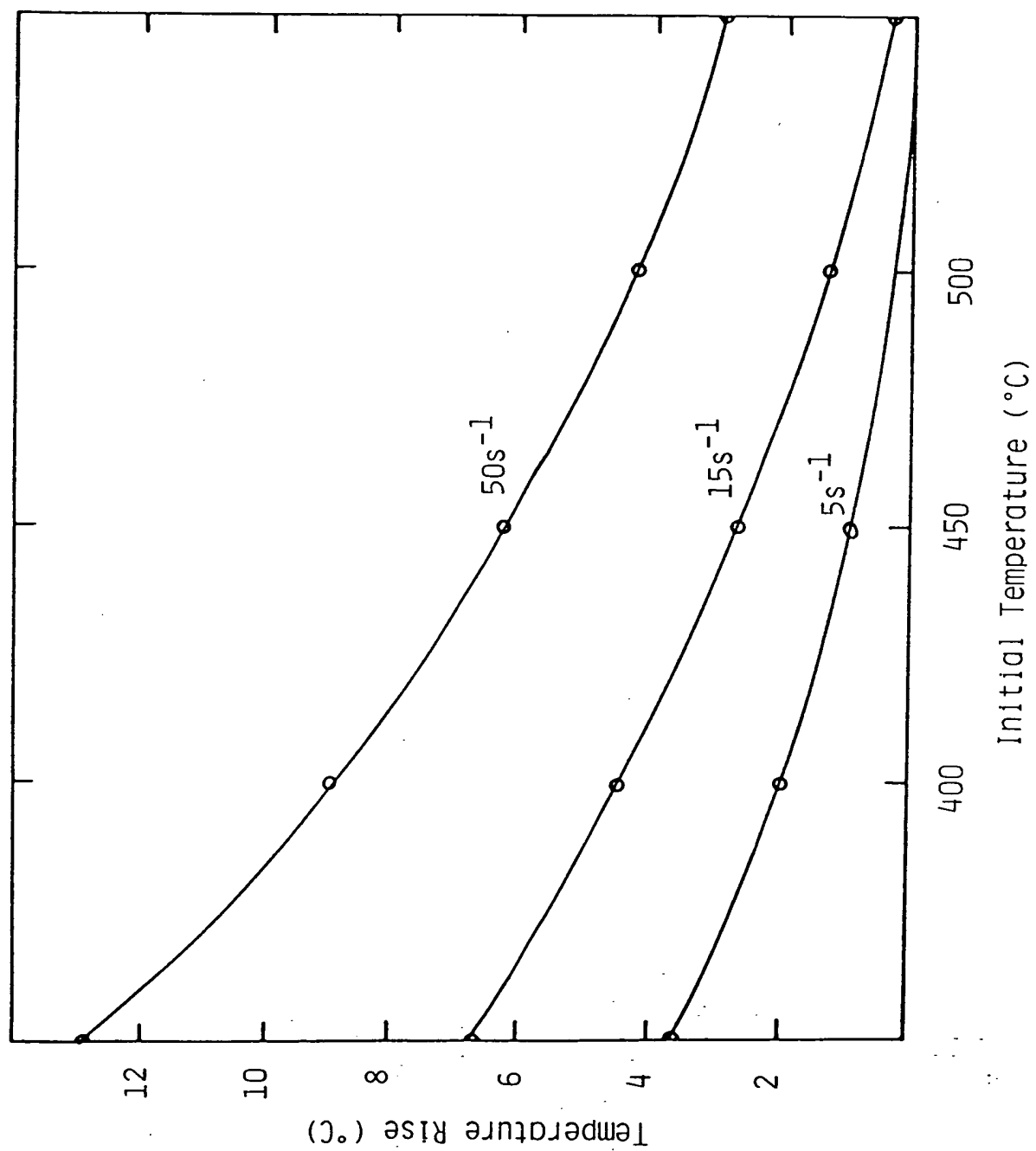


Figure 5.3. Temperature Rise During Torsion

Table 5.2.

Torque (Nm)	Twist Rate (rad. ⁻¹)	Initial Temp. °C	Effective Stress (MNm ⁻²)	Effective Strain-Rate (s ⁻¹)
6.8950	1.00	350	35.20	0.29
5.3821	1.05	400	27.37	0.30
3.7876	1.00	450	21.19	0.29
2.1175	1.10	500	16.96	0.32
3.6183	0.20	400	21.75	0.06
6.4008	0.22	350	28.94	0.06
5.3806	9.14	400	35.94	2.64
7.9122	9.43	350	44.76	2.72
4.2365	9.09	450	28.94	2.62
3.2314	9.52	500	23.60	2.75
8.0407	31.29	350	50.08	9.03
7.4739	93.30	350	55.00	26.93
5.9701	93.30	400	45.92	26.93
6.2875	30.10	400	40.98	8.69
4.7564	30.62	450	33.73	8.84
4.7987	93.78	450	38.37	27.07
3.6878	30.57	500	27.78	8.83
4.0355	93.34	550	26.92	26.66
3.1664	30.14	550	22.97	8.70
2.5890	9.86	550	19.41	2.85
1.9663	0.70	550	12.54	0.20

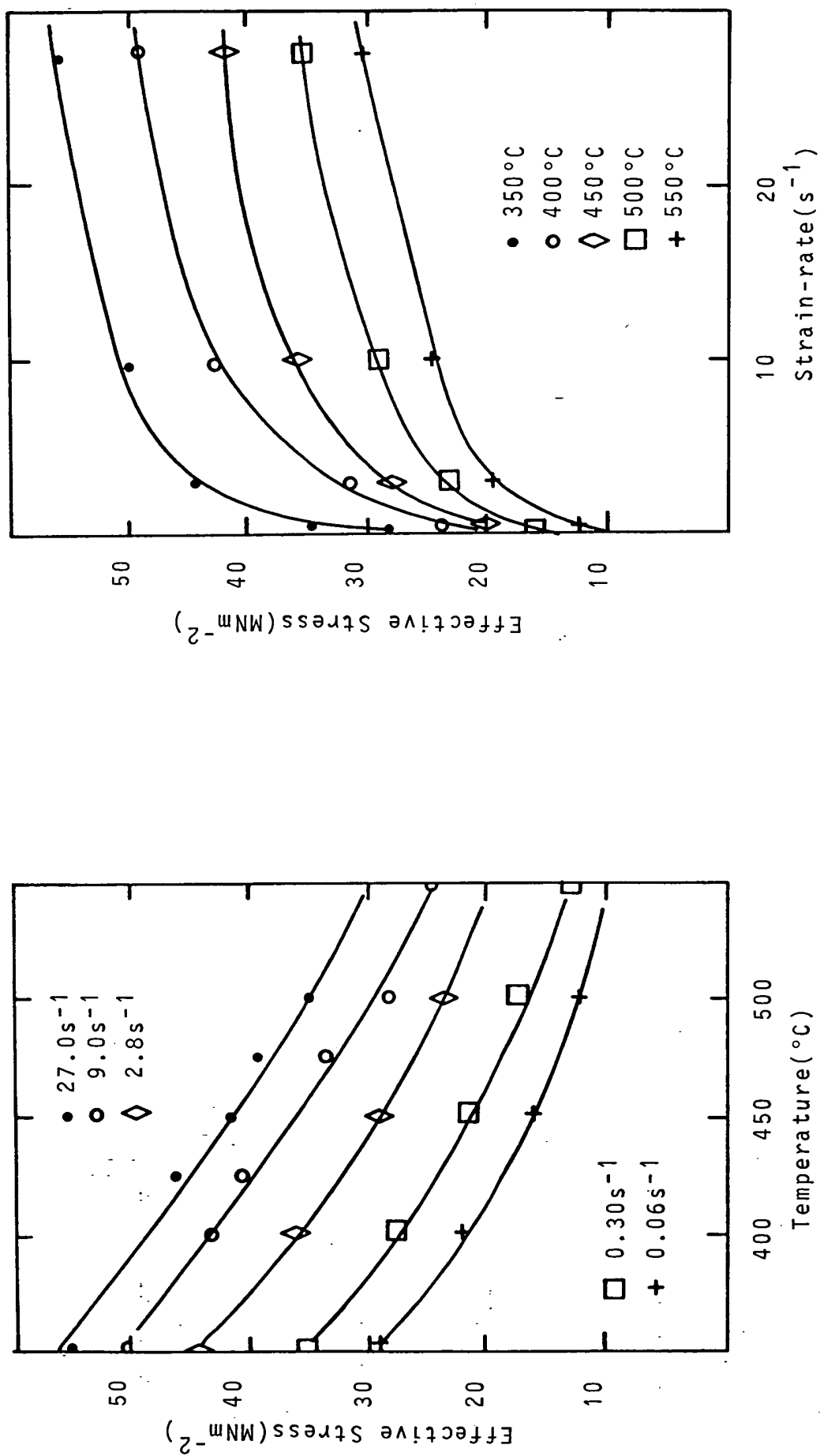


Figure 5.4. The Effects of Strain-Rate and Temperature on Flow Stress (Zener-Hollomon)

This method of obtaining flow stress from torsion data has been used successfully for numerous aluminium alloys^(6-9,11,13,14) and ferrous materials^(12,100). The main drawback of this technique is that to obtain an accurate solution, 25 temperature/strain-rate combinations are required to allow the hot working constants to be calculated for subsequent flow stress evaluation. The strain dependence of the flow stress is also difficult to establish. Using this method, Vierod⁽¹¹⁾ suggested minimising the developed torque at specific homologous strains to calculate this dependence. However, even with the improved computer data collection and analysis, this proved to be tedious and moreover resulted in considerable scatter of the results. In view of the large deviation in the results and frequent minimisation failure, this particular approach to obtain a strain dependence for AA6063 was abandoned.

5.2.2.2. Effective Radius

An alternative method of evaluating flow stress was discussed in Section 3.2.2. This technique has the advantage of calculating the flow stress for any strain within the test, from the results of one torque/twist curve. The results from the application of this method to the torsion data of AA6063 are listed in Table (5.3.). In Figure (5.5.) these results are plotted demonstrating clearly the influence of temperature and strain-rate on the hot working flow stress.

The main advantage of this method is that with suitable programming, simultaneous conversion of the data can be made as the test is conducted, and as the results correspond to a radius within the

Table 5.3.

Torque (Nm)	Twist Rate (rad. ⁻¹)	Initial Temp. °C	Effective Stress (MNm ⁻²)	Effective Strain Rate (s ⁻¹)
6.8950	1.00	350	45.60	0.2098
5.3821	1.05	400	35.60	0.2211
3.7876	1.00	450	25.10	0.2080
2.1175	1.10	500	13.90	0.2290
3.6183	0.20	400	23.90	0.0409
6.4008	0.22	350	42.30	0.0470
5.3806	9.14	400	35.60	1.9092
7.9122	9.43	350	52.40	1.9690
4.2365	9.09	450	28.00	1.9007
3.2314	9.52	500	21.40	1.9941
8.0407	31.29	350	53.20	6.5400
7.4739	93.30	350	49.40	19.5000
5.9701	93.30	400	39.50	19.5000
6.2875	30.10	400	41.70	6.2929
4.7564	30.62	450	31.50	6.4004
4.7987	93.78	450	31.80	19.5616
3.6878	30.57	500	24.40	6.3846
4.0355	93.34	550	26.70	19.3300
3.1664	30.14	550	20.90	6.2974
2.5890	9.86	550	17.13	2.0616
1.9663	0.70	550	13.00	0.1469

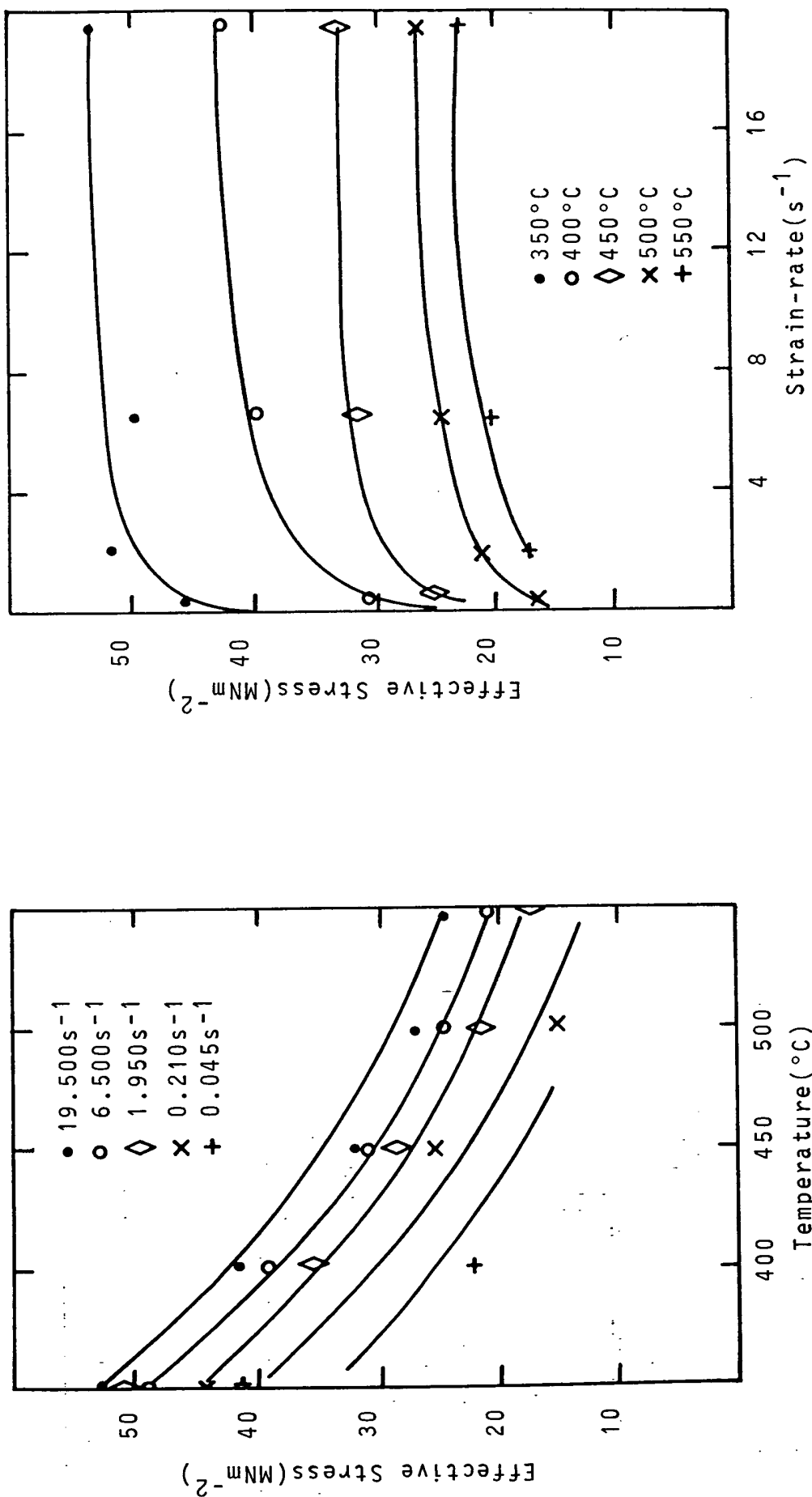


Figure 5.5. The Effects of Strain-Rate and Temperature on Flow Stress (Effective Radius Analysis)

specimen, the influence of the inhomogenous flow at the specimen surface, due to strain roughening, is eliminated.

5.2.3. Flow Stress Characteristics

It was shown in the previous sections that the process parameters, temperature and strain-rate have a pronounced effect on the flow stress at high homologous temperatures. This phenomenon is most conveniently explained by considering the thermally activated events obtaining in elevated temperature deformation.

The main softening mechanism in AA6063, at hot working temperatures, is dynamic recovery, and it is the ease by which recovery can occur that directly governs the flow stress characteristics. The two most obvious factors are the process parameters; temperature and strain-rate.

An increase in temperature reduces the flow stress as thermally-activated cross-slip and climb of dislocations increase the rate of recovery. A rise in strain-rate increases the dislocation generation rate and effectively reduces the time for the recovery process, resulting in a less recovered sub-structure and higher flow stress. The opposing effects of temperature and strain-rate are most conveniently combined into a single parameter 'Z', the temperature compensated strain-rate. Previous workers^(7,11,14) have shown that 'Z' can be used to predict a specific sub-structure or flow stress for any combination of temperature and strain-rate.

Apart from temperature and strain-rate, another important factor governing recovery and, therefore, flow stress is the material itself. The rate of dynamic recovery will depend on the nature of the metal (e.g. pure metal, solid solution, nature of precipitate and particle distribution, etc.) and so will ultimately determine the flow stress characteristics described by the temperature/strain-rate relationships.

In the present study, the effect of temperature and strain-rate on steady state flow stress characteristics were determined for AA6063 using the hot torsion test.

Flow stress data relating dependence on strain-rate and temperature are most conveniently represented in the form of a constitutive equation. The nature of the torque minimisation conversion method produces, by necessity, a constitutive law. However, the data from the effective radius analysis must be subjected to further mathematical treatment in order to establish a constitutive law. The constitutive equations used to represent the flow stress data are discussed in the following sections.

5.2.4. Constitutive Equations

The connection between the externally measureable parameters, such as flow stress, strain-rate and temperature are called constitutive relationships. These relationships can also be extended to include an internal variable which allows the effect of the metallurgical state of the material, on the flow stress, to be included.

In the present work, three forms of constitutive equation were assessed using the flow stress data from hot torsion testing of AA6063. The first equation relates the steady state flow stress with temperature and strain-rate. The second two equations also relate steady state flow stress with temperature and strain-rate but in addition describes the transient behaviour of the flow stress from the point of yielding to steady state conditions. These equations will now be discussed in turn.

5.2.4.1. Steady State Constitutive Law

(Zener-Hollomon Equation)

It is generally accepted that in the hot working range, where $T > 0.5T_m$ (T_m = melting point (K)), the deformation mechanisms are an extension of those occurring in creep. Recognition of this fact led Sellars et al⁽¹¹³⁾ to extend the use of an equation originally proposed for creep by Garafalo⁽¹⁰¹⁾ to include the hot working regime. This equation is generally rearranged into a form which gives a temperature compensated strain-rate parameter 'Z':-

$$Z = \dot{\epsilon} \exp\left(\frac{\Delta H}{RT}\right) = A (\sinh(\alpha\sigma))^n \quad -(5.2.).$$

where: ΔH = Activation energy of the rate controlling mechanism

α, n, A = Hot working material constants

R = Universal gas constant.

The experimental data from the hot torsion tests on AA6063 were used to establish the four material constants A , α , n , and ΔH in Equation

(5.2.), following the method described in Section 3.2.1. The constants obtained from this analysis, describing steady state flow stress are listed in Table (5.4.).

ΔH	= 141550	J/mole
α	= 0.04	
n	= 5.385	
$\ln A$	= 22.5	
cc	= 0.9859	

Table (5.4.): Hot Working Constants for AA6063
(in the homogenised condition)

For the purpose of this study, the steady state is defined as 'the point at which the derivative of the flow with respect to strain is zero as a result of the recovery processes balancing dislocation storage rate'.

The presence of an Arrhenius term in Equation (5.2.) implies that hot working is a thermally activated process where ΔH is equal to the activation energy of the rate controlling mechanism. The value of 141550 J/mole obtained for AA6063 is close to that reported for self diffusion of vacancies. This confirms that dynamic recovery is the softening mechanism operative during deformation. Therefore we can say that the magnitude of ΔH represents the size of the barrier preventing atomic rearrangement. However, it must be noted that other mechanisms may be occurring concurrently whose energy barriers are lower than ΔH .

The remaining empirical constants A , α and n do not have a simple physical interpretation, although some attempts have been made⁽⁵⁰⁾.

Comparing the results for AA6063 established in this study with other workers^(7,11,14) it appears that ' ΔH ', and ' A ' can be regarded as the material constants related to the diffusion characteristics of the matrix and hence the rate controlling mechanism. Whilst the constants ' α ' and ' n ' describe the flow stress characteristics. The term ' α ' relates to the saturation value of the slowest evolving state parameter and the constant ' n ' can be interpreted as the inverse of the temperature compensated strain-rate sensitivity.

Assessment of Equation (5.2.) ability to predict steady state flow stress is made in Figure (5.6.). Here, the flow stress calculated using the constants listed in Table (5.4.) are plotted against the actual (effective radius) flow stress. It can be seen that at low stress values i.e. low ' Z ', there is an excellent correlation between the experimental and predicted values. At high ' Z ' conditions there is an increase in the deviation of the results. However, the deviation remains about the line of unity gradient, passing through the origin, relating σ_{test} and $\sigma_{\text{prediction}}$. The decreased correlation at high stress levels might be explained in terms of the commonly made definition of ' α ' as the reciprocal stress at which the temperature compensated strain-rate changes from power to exponential stress dependence. For AA6063, in the present trials, this point was $1/\alpha = 25\text{MNm}^{-2}$ which approximately corresponds to the point of departure of good correlation shown in Figure (5.6.). This does not imply failure of the Zener-Hollomon equation to predict high ' Z ' stress values. The reason for the decreased accuracy of the equation (5.2.) is that the low strength AlMgSi alloy used in the present research resulted in the majority of the data points laying in the low stress region. This effectively weighted the data to produce hot

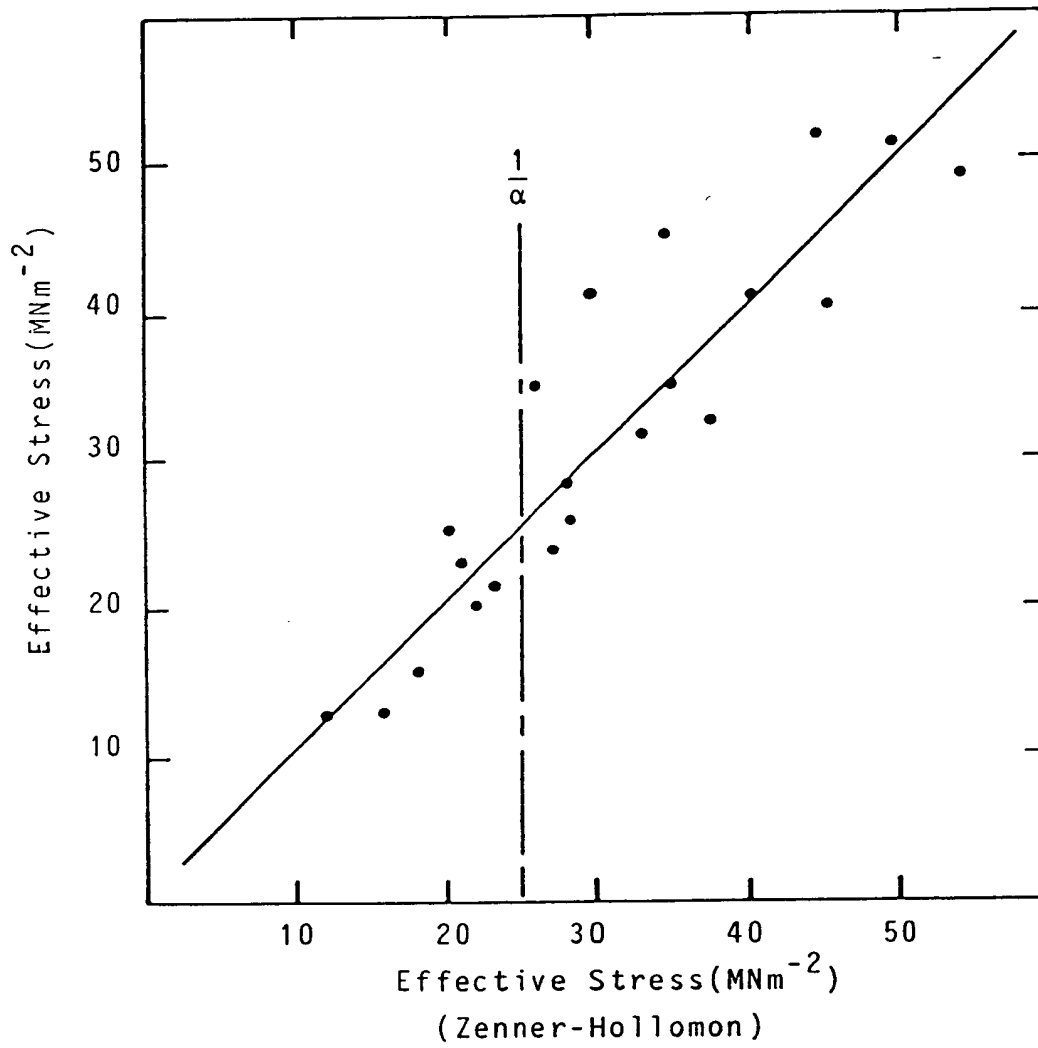


Figure 5.6. Test Flow Stress Vs. Predicted Flow Stress
(Zener-Hollomon)

working constants with a low stress (low Z) regime bias.

5.2.4.2. Strain Based Constitutive Law

(Anand)

The analysis of the flow stress dependence on the plastic strain-rate and absolute temperature at steady state conditions, using the Zener-Hollomon equation, is widely used and has proved to be of considerable use. However, there is increasing demand for improved constitutive relationships of a phenomenological nature. Guided by internal-variable theories⁽¹⁰⁸⁻¹¹¹⁾, a constitutive equation of the type proposed by Anand⁽⁹⁰⁾ will be explored in this section to assess its applicability to deformation of AA6063 at high homologous temperatures.

It was shown in Chapter 2 that for uniaxial loading of constant sign and assuming isotropic hardening, it was possible to divide the plastic behaviour into two mechanistic steps: The flow stress dependence on strain-rate, temperature and prevailing structure, and the evolution of structure with strain:-

$$\sigma = \sigma(s_i, \dot{\epsilon}, T) \quad -(5.3.a)$$

$$\frac{ds_i}{d\epsilon} = (s_i, \dot{\epsilon}, T) \quad -(5.3.b)$$

Here, s_i consists of a list of internal variables characterising the resistance to plastic flow caused by the internal microstructural state of the material.

At this initial stage of analysis it is prudent to combine the list of internal variables, to contain a single, non-zero, positive valued scalar with the dimensions of stress. (At this point no attempt will be made to assign any physical significance to the parameter.) It is clearly an over simplification that a single scalar can characterise the initial structure. However, under the assumptions made, the approach is worth pursuing.

The kinetic equation adopted in this analysis is one suggested by Lee and Zaverl⁽⁹²⁾ which has the familiar form of an Arrhenius-type rate equation:-

$$\dot{\epsilon} = A \exp\left(\frac{-\Delta H}{RT}\right) (\sigma/s)^{1/m} \quad -(5.4.)$$

In this equation $\sigma < s$.

In addition to the kinetic equation, an evolution equation is required for a complete description. In this analysis a simple form of evolution equation for 's' proposed by Anand⁽⁹⁰⁾ will be adopted:-

$$\dot{s} = h_0 \left(1 - \frac{s}{s^s}\right) \dot{\epsilon} \quad -(5.5.a)$$

$$\text{where:} \quad s^s = \hat{s} \left(\frac{\dot{\epsilon}}{A} \exp\left(\frac{\Delta H}{RT}\right) \right)^n \quad -(5.5.b)$$

and h_0 , $\hat{s}$, and n are material parameters.

' s^s ' corresponds to the saturation value of s at $\dot{s}=0$ defined by a given temperature/strain-rate pair.

An important feature of Equation (5.5.a) is that it is composed of a

constant hardening rate ' h_0 ' and a variable softening rate ' $h_0 s/s^S$ ' permitting the influence of temperature and strain-rate to be incorporated.

When plastic flow has developed- at a fixed temperature and strain-rate - equation (5.4.) implies that the internal variable ' s ' is proportional to the external variable ' σ ':-

$$\begin{aligned} \sigma &= s \cdot c & -(5.6.a) \\ \text{where: } c &= \frac{\dot{\epsilon}}{A} \exp\left(\frac{\Delta H}{RT}\right)^m & -(5.6.b) \end{aligned}$$

Assuming that the temperature rise from the point of yielding to peak stress is negligible, the isothermal constant strain-rate conditions allow Equation (5.5a) to be written:-

$$\begin{aligned} \frac{d\sigma}{d\epsilon} &= ch_0 \left(1 - \frac{\sigma}{\sigma^S}\right) & -(5.7.a) \\ \text{where: } \sigma^S &= s^S c & -(5.7.b) \end{aligned}$$

Intergrating Equation (5.7.a) results in the Voce⁽¹¹⁵⁾ stress/strain relationship:-

$$\begin{aligned} \sigma &= \sigma^S - (\sigma^S - \sigma_0) \exp(-\epsilon/\epsilon^S) & -(5.8.a) \\ \text{where: } \sigma_0 &= s_0 c & -(5.8.b) \\ \text{and, } \epsilon^S &= \sigma^S / (ch_0) & -(5.8.c) \end{aligned}$$

The value of strain (ϵ) in Equation (5.8.a) consists solely of the plastic component (i.e. total strain-elastic strain).

The calculated flow stress/strain data from the torsion tests, using effective radius conversion methods, were applied to the above kinetic

and evolution equation using least-square methods.

In order to calculate the constants of the constitutive equation the data must be regressed subject to Equation (5.7.a). This requires the evaluation of $d\sigma/d\epsilon$ at σ values from yielding to steady state (or peak stress). Because the data from the torsion tests were collected via a computer, this was achieved simply by applying standard central difference formula to the digitised data points.

Typical plots of $d\sigma/d\epsilon$ Vs σ together with the stress/strain curve from the experimental data can be seen in Figure (5.7.a-d) for various temperature/strain-rate conditions. These plots demonstrate that the data has the required form to fit Equation (5.7.a). From the regression analysis, values of σ and ch_0 are obtained for each test condition. These are listed in Table (5.5.). Using the σ^s values, Equation (5.8.a) can be extrapolated to find σ_0 . The material parameters A , ΔH , $\hat{s}$ and m' are then determined by minimising Equation (5.7.b). Then, using the values obtained for A , ΔH and σ_0 , non-linear, least-square estimates can be made for s_0 and m using 5.4. The temperature variation of m and s_0 is determined by repeating the procedure at various temperatures. Finally the value of h_0 is obtained by averaging the calculated ' ch_0 ' values.

The data collection and analysis was performed using programs⁽¹¹⁴⁾ written for the IBM PC with A/D capabilities. The correctly formatted data files were then transferred to a mainframe computer in order to establish the material constants of the proposed constitutive law. The mainframe computer program, written by Bianchi⁽⁹³⁾, used non-linear estimation techniques using NAG library routines.

σ^S	σ_0	ch_0	---	ϵ	T	$\epsilon\%$
.6314E+02	.1782E+02	.2717E+03	.2718E-03	.2098E+00	.3500E+03	.3253E-01
.4236E+02	.1477E+02	.2905E+03	.1943E-03	.2211E+00	.4000E+03	.2042E-01
.2794E+02	.1156E+02	.2436E+03	.1065E-03	.2080E+00	.4500E+03	.1606E-01
.1446E+02	.1174E+02	.6776E+02	.1522E-04	.2295E+00	.5000E+03	.2989E-01
.2559E+02	.1272E+02	.4733E+03	.2582E-04	.4090E-01	.4000E+03	.7570E-02
.4779E+02	.9678E+01	.7908E+03	.6694E-04	.4700E-01	.3500E+03	.8460E-02
.5352E+02	.1311E+02	.5759E+02	.6650E-03	.1909E+01	.4000E+03	.1301E+00
.5640E+02	.3197E+02	.9470E+02	.2627E-03	.1969E+01	.3500E+03	.8338E-01
.4494E+02	.1292E+02	.2887E+02	.3916E-03	.1901E+01	.4500E+03	.2179E+00
.2584E+02	.9977E+01	.3134E+02	.2333E-03	.1994E+01	.5000E+03	.1154E+00
.9418E+02	.2020E+02	.3274E+02	.1634E-02	.6540E+01	.3500E+03	.4027E+00
.2575E+03	.3566E+02	.2059E+02	.5942E-02	.1950E+02	.3500E+03	.1751E+01
.4028E+03	.5331E+02	.6770E+01	.2288E-02	.1950E+02	.4000E+03	.8329E+01
.7587E+02	.1516E+02	.2803E+02	.1463E-02	.6293E+01	.4000E+03	.3789E+00
.6910E+02	.1015E+02	.1731E+02	.1150E-02	.6400E+01	.4500E+03	.5587E+00
.7329E+02	.1153E+02	.9528E+01	.2069E-02	.1956E+02	.4500E+03	.1077E+01
.3200E+02	.9441E+01	.1691E+02	.5068E-03	.6385E+01	.5000E+03	.2650E+00
.4317E+02	.1024E+02	.6636E+01	.1059E-02	.1933E+02	.5500E+03	.9108E+00
.2891E+02	.7782E+01	.1246E+02	.4648E-03	.6297E+01	.5500E+03	.3248E+00
.2092E+02	.8729E+01	.2976E+02	.2386E-03	.2062E+01	.5500E+03	.9840E-01
.1360E+02	.9776E+01	.1640E+03	.3443E-04	.1469E+00	.5500E+03	.1161E-01

Table 5.5. Data Output From Anand Treatment

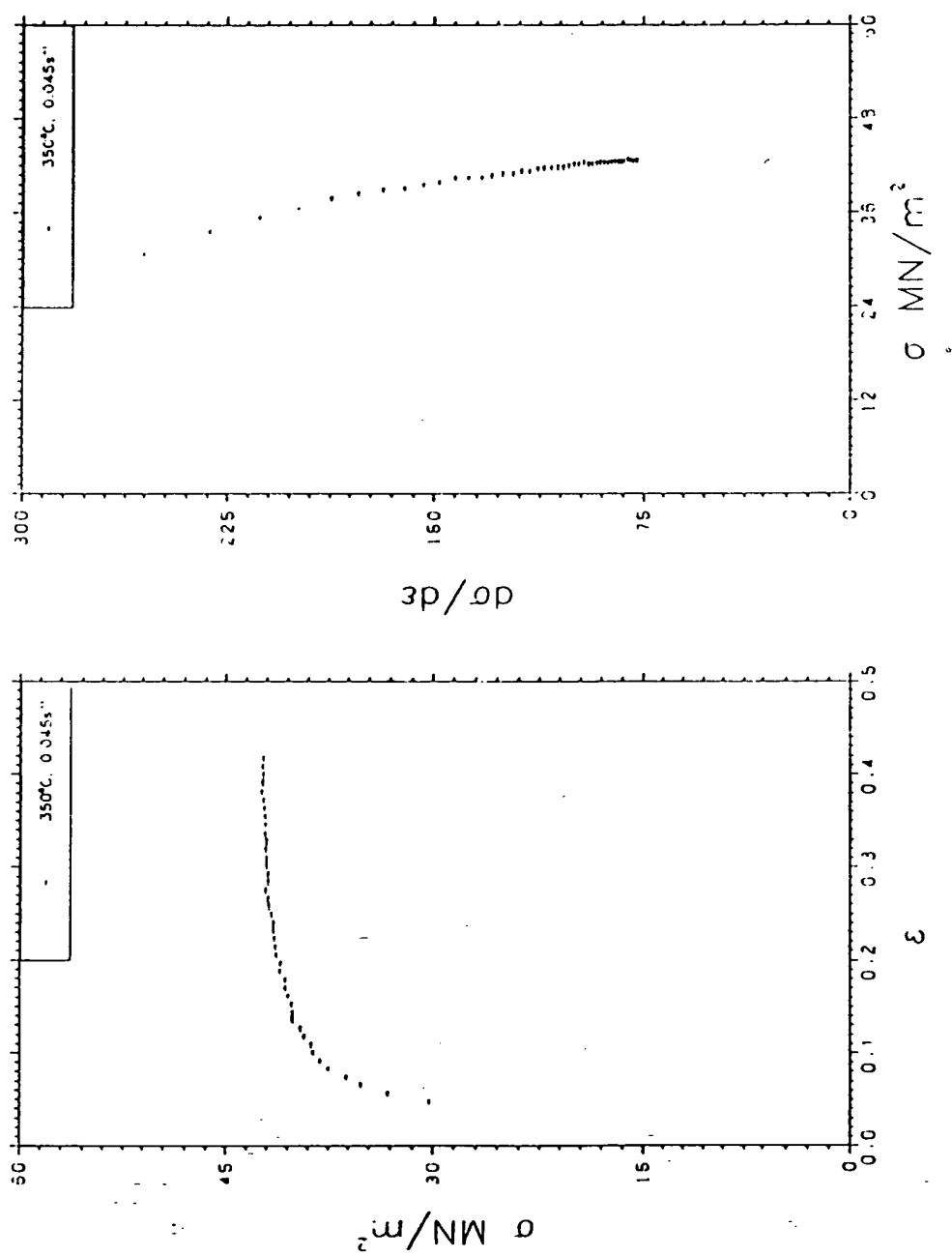


Figure 5.7. (a) Test Data Output

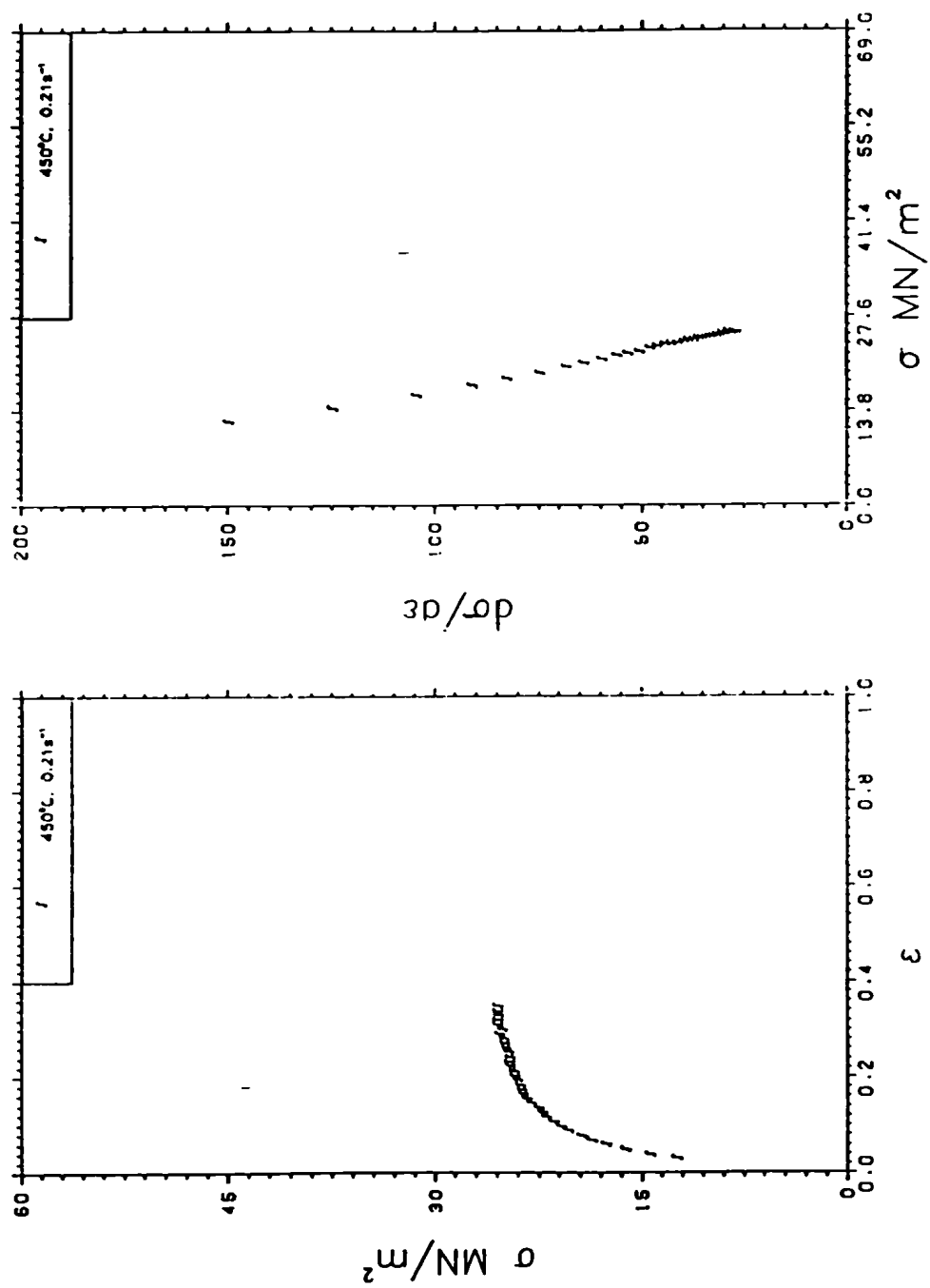


Figure 5.7. (b) Test Data Output

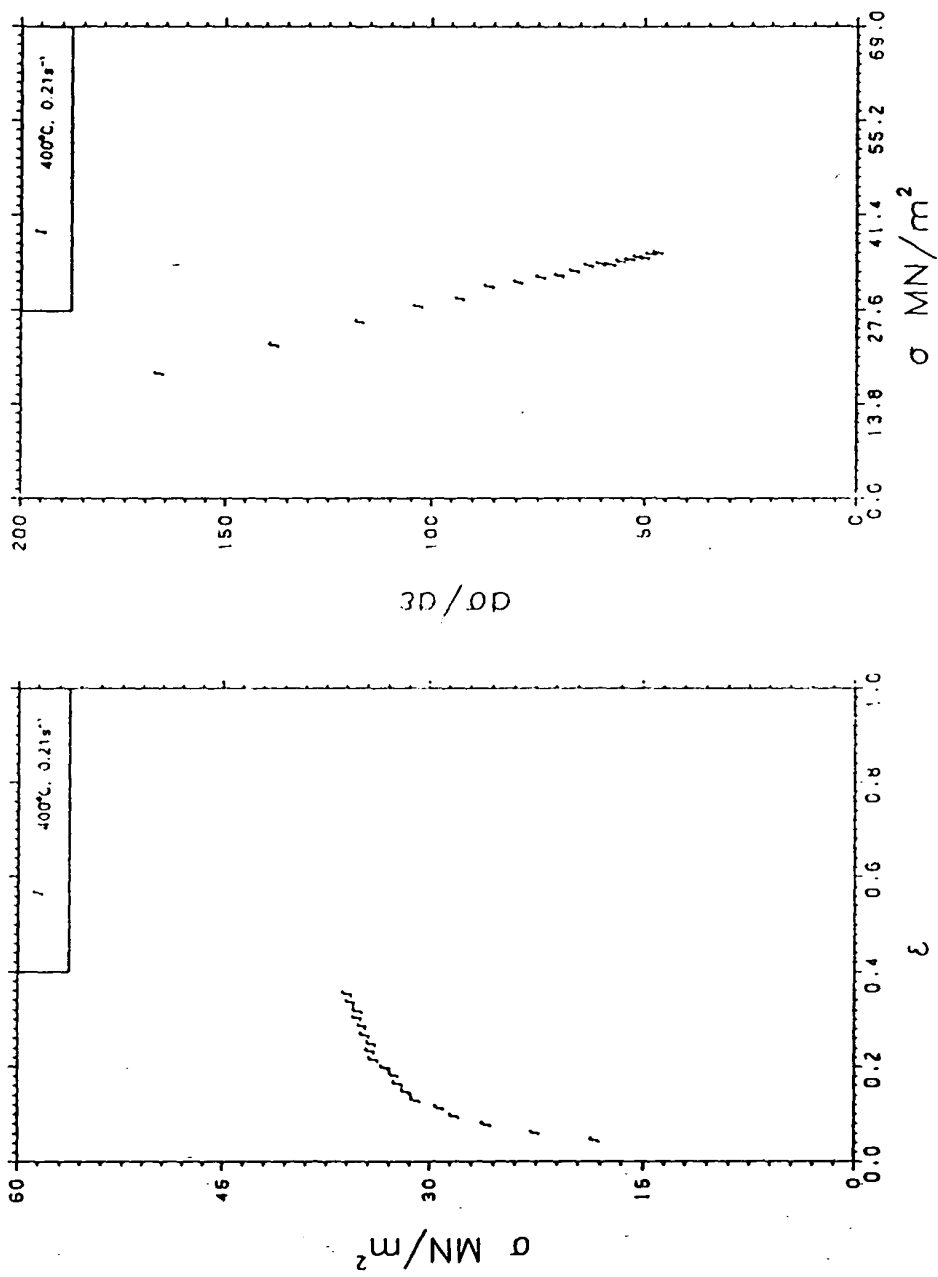


Figure 5.7. (c) Test Data Output

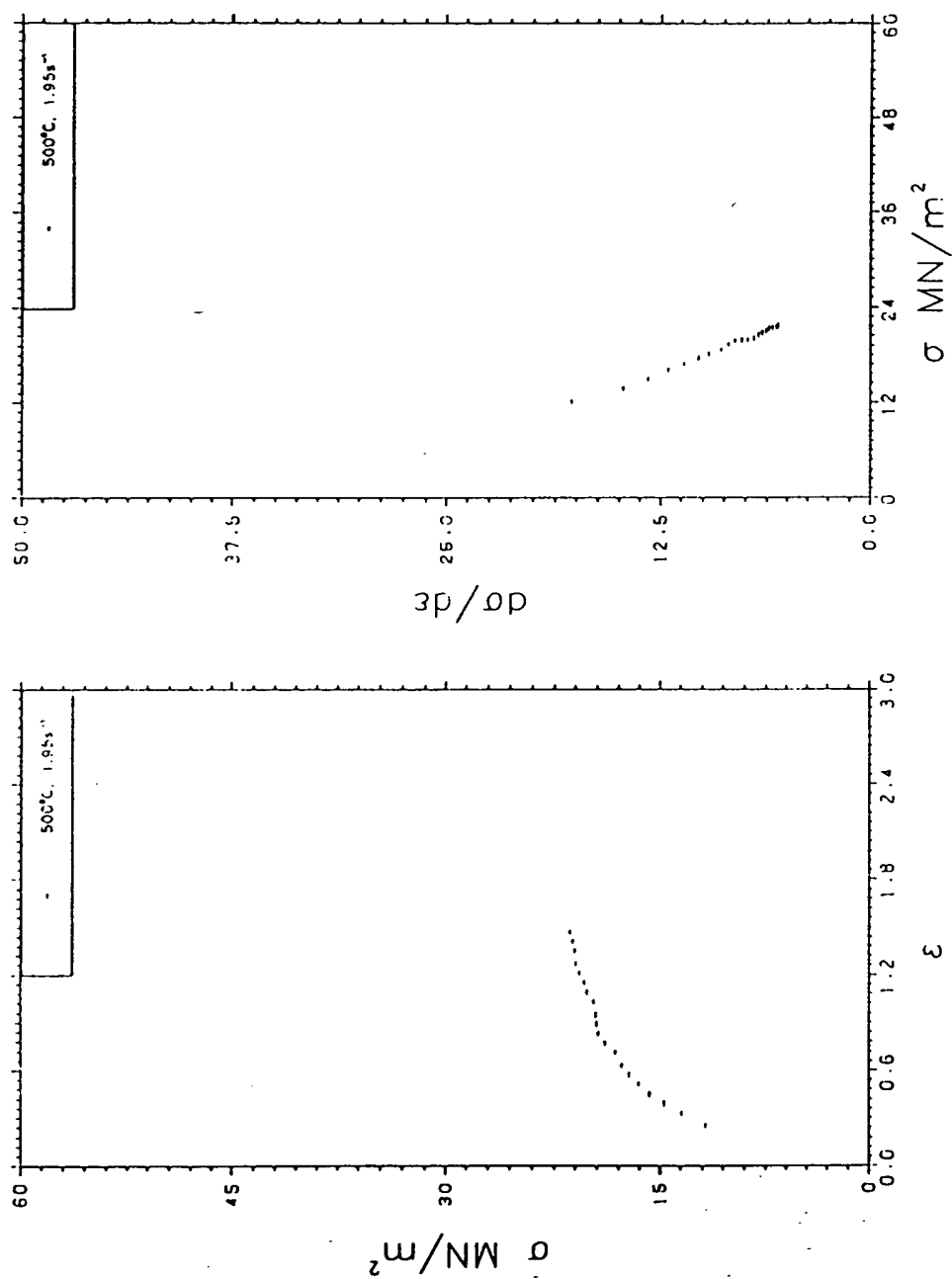


Figure 5.7. (d) Test Data Output

The material constants obtained from the above analysis for the kinetic and evolution equations are listed in Table (5.6.). These constants may now be used to predict steady state flow stress and describe transient behaviour from yielding to steady state at defined temperature and strain-rate conditions.

Table (5.6.) Material Constants For Anand Equation

$\Delta H =$	139500
$h_o =$	10.0757
$A =$	39.6960
$m =$	0.1322
$n =$	0.1393
$\hat{s} =$	0.1355
$s_o =$	0.8483
$\epsilon_o =$	0.0412

Interpretation of the physical significance of the material constants is difficult at this stage, as there is insufficient data to assess the effects of the alloy's content, thermomechanical history and testing methods. However, some tentative suggestions can be made from the observation made on the current tests.

The value for ΔH of 139500 J/mole is consistent with the value obtained in the Zener-Hollomon treatment and represents the activation energy of the rate controlling mechanism. (In this event the value of the ΔH is close to that reported for self diffusion of vacancies and so is believed to be dynamic recovery). 'A' appears to be related to the diffusion characteristics of the metal, whilst ' h_o ' is a measure of the work hardening rate. The indices ' m ' and ' n ' are sensitivity constants for the temperature compensated strain-rate and the evolution of structure respectively.

The most important feature in the Anand model is the introduction of the 'state' variable ' s ' and its associated constants s^s , s_0 , and $\dot{s}$, thus allowing the internal 'state' of the material to be incorporated. If such a state variable exists, then it can be identified with some strength determining property of the dislocation sub-structure which is developed during the course of the plastic deformation. One such property is the material's hardness. Using a modified Anand approach, Sample and Lalli⁽⁹⁸⁾ related the state variable to the micro-hardness with considerable success for commercial purity aluminium. However, no such correlation could be obtained in the present research. The principal reason being that AA6063 is a heat-treatable alloy and therefore solution strengthening disguised the hardening component due to the dislocation arrangement. This clearly casts doubts as to whether one state variable is sufficient to describe flow stress behaviour of heat-treatable alloys.

In Figure (5.8.) the experimental flow stress is plotted against the predicted flow stress using the Anand model. It is clear that at low stress levels (low ' Z ') there is a good estimate of the steady state flow stress. However, as the stress level is increased there is an apparent failure of the proposed Anand model for AA6063. Obviously, the inability to predict the steady state flow stress at high ' Z ' conditions also implies failure of the transient description of flow stress with strain. This feature is demonstrated in Figure (5.9.) for high, medium and low ' Z ' conditions.

The failure of the Anand treatment at high stress levels can be attributed to two main sources. Firstly, the use of a power law form

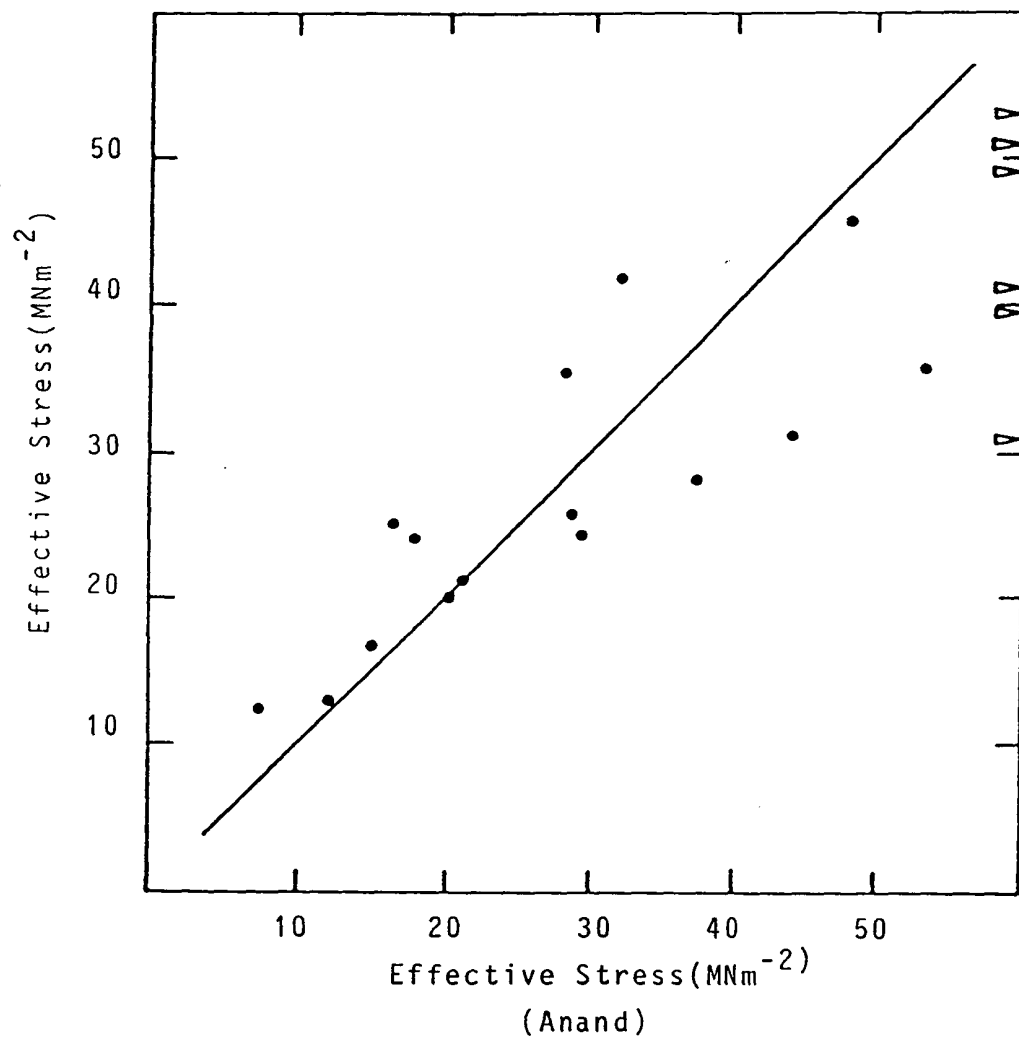


Figure 5.8. Test Flow Stress Vs. Predicted Flow Stress (Anand)

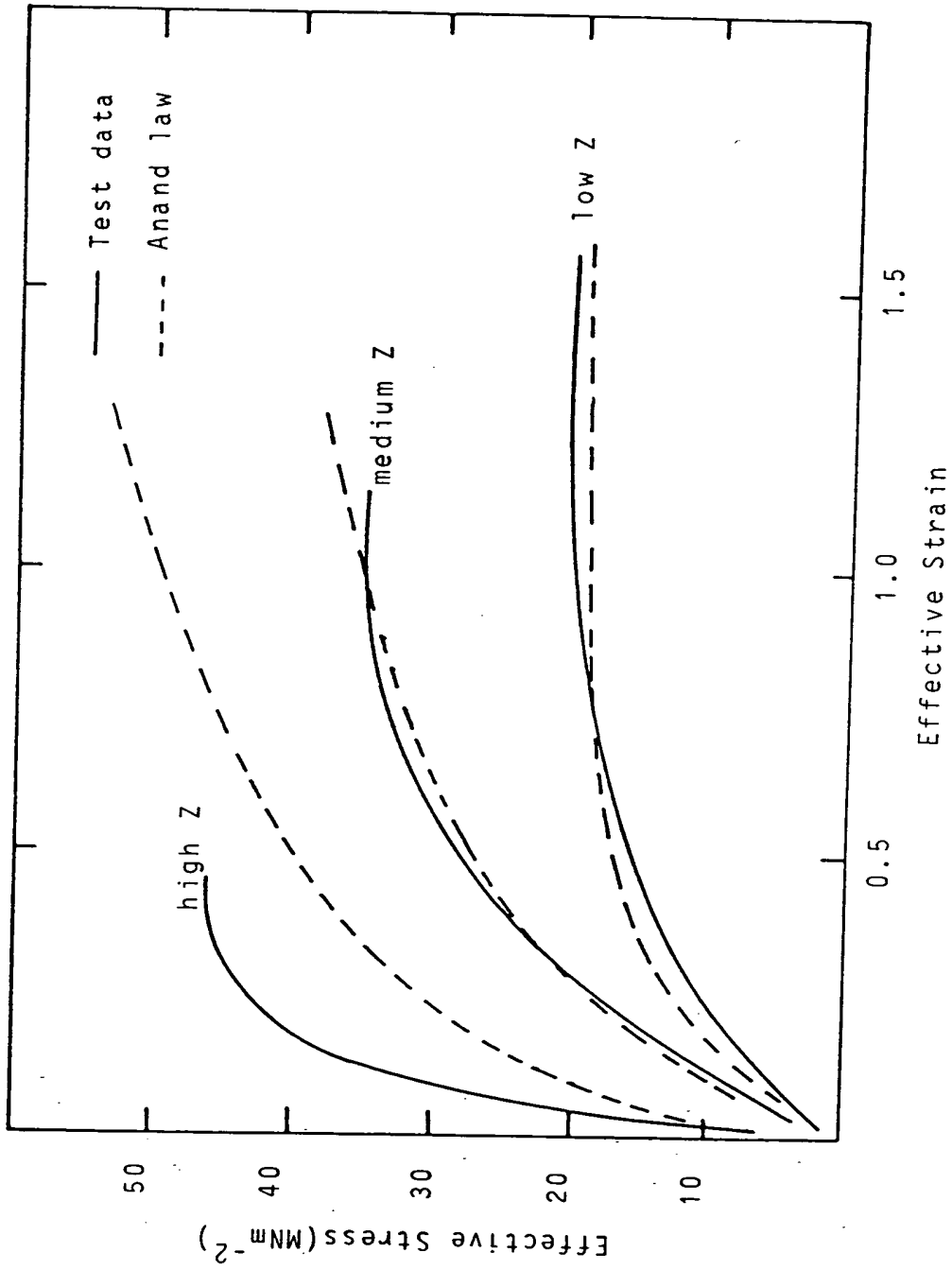


Figure 5.9. Experimental and Predicted Flow Stress Curves Using the Anand Model.

for the kinetic equation has been shown⁽⁵⁰⁾ to be inadmissible at high stress levels. This feature was also demonstrated for the Zener-Hollomon treatment in Section 5.2.4.1.

The second reason for the failure of the Anand model is most conveniently explained by considering the dynamic recovery mechanism. As each specimen is placed under load, there is an athermal work hardening due to the generation and interaction of dislocations. The resultant stress value is obtained from the balance of work hardening and the concurrent operation of dynamic recovery. Because dynamic recovery is a function of the strain-rate, temperature and the dislocation density, as strain increases, the dynamic recovery increases due to a rise in dislocation density, until steady state is reached where generation and annihilation of dislocations are in balance. In the Anand treatment, the evolution equation is fitted to the strain hardening values of the experimental data.

Because this equation is fitted to the transient portion of the data, where the recovery has not yet fully developed, an under estimate of the contribution is made. This feature is shown schematically in Figure (5.10.). This results in an over estimate of the steady state flow stress which becomes greater at high 'Z' conditions.

It is clear that the constitutive equation proposed by Anand requires further development in order to increase its validity range. In the following section, an alternative kinetic equation is proposed, whilst retaining the same evolution equation.

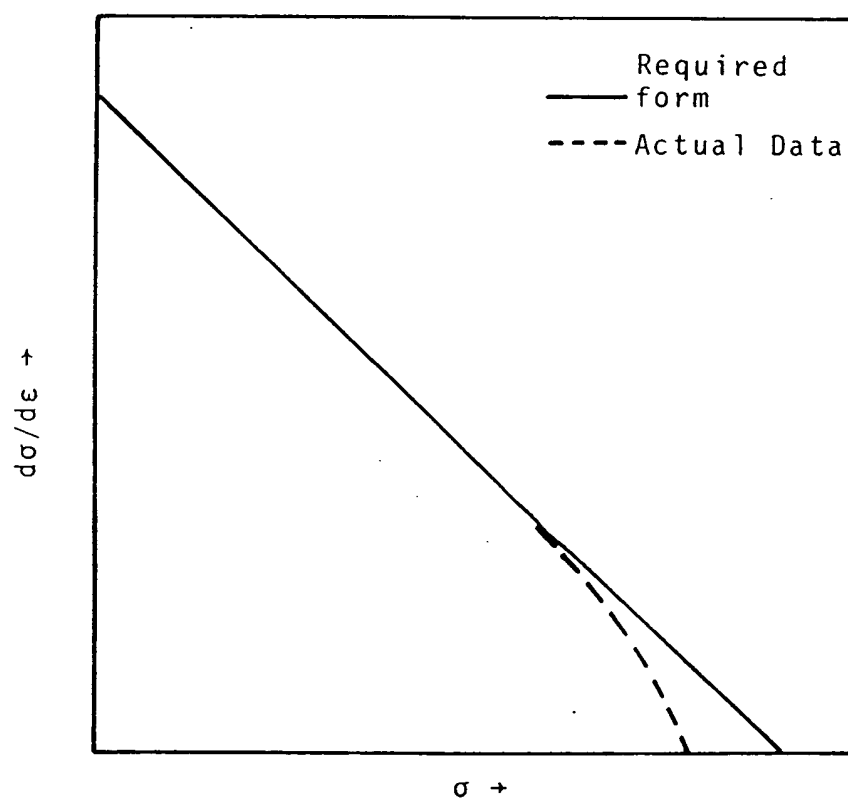


Figure 5.10. Schematic of $d\sigma/d\epsilon$ Vs σ

5.2.4.3. Strain Based Constitutive Law

(Hybrid Model)

It was shown in the previous section that the Anand model was unable to produce good estimates of the flow stress at high 'Z' conditions, and it was proposed that this was partly due to the departure from power law dependence. In order to pursue this hypothesis, a hyperbolic sine kinetic equation, similar to that of Zener-Hollomon was adopted:-

$$\dot{\epsilon} = A \sinh^{-1} \left(\frac{\sigma}{s} \right)^{1/m''} \cdot \exp \left(- \frac{\Delta H}{RT} \right) \quad -(5.9)$$

The constants in Equation (5.9.) take their usual meaning and the evolution equation for the structure parameter 's' is again that proposed by Anand (Equation 5.5.).

An assessment of the hybrid (Zener-Hollomon/Anand) model was made using the same data files and computational procedures as those used for the Anand treatment. The only difference between the Anand model and the one proposed here is in the test constant 'c' which is replaced by:-

$$c' = \sinh^{-1} \left(\frac{\dot{\epsilon}}{A} \exp \left(\frac{\Delta H}{RT} \right) \right)^{m''} \quad -(5.10.)$$

The constants obtained from this analysis are listed in Table (5.7.).

Table 5.7. Material Constants for Hybrid Equation

ΔH	=	142 200
h_0	=	10.64
A	=	41.26
m''	=	0.59
n	=	0.13934
$\hat{s}$	=	0.13552
s_0	=	0.8480

These constants are in good agreement with those obtained for the previous model, with most noticeably the value for activation energy, which is also consistent with that obtained from the Zener-Hollomon equation.

These constants can now be used in conjunction with Equation (5.9.) to predict the flow stress at various strain, temperature and strain-rate combinations. In Figure (5.11.) a plot of experimental flow stress against predicted flow stress, using the Hybrid constitutive law, is shown. The graph demonstrates that there is an improvement in the correlation due to the increased accuracy of flow stress estimates. However, there is still some evidence of a decrease in correlation at high stress levels but to a much lesser extent than that of the Anand model.

In addition to the improved accuracy of the predicted steady state flow stress, there is an improvement in the transient description of the flow stress from the yield point to the steady state. This is clearly demonstrated in Figure (5.12.) for high, medium and low 'Z' conditions. The stress/strain curves at the medium and low 'Z' conditions are reproduced with reasonable accuracy using the Hybrid model, but at high 'Z' there is still considerable deviation.

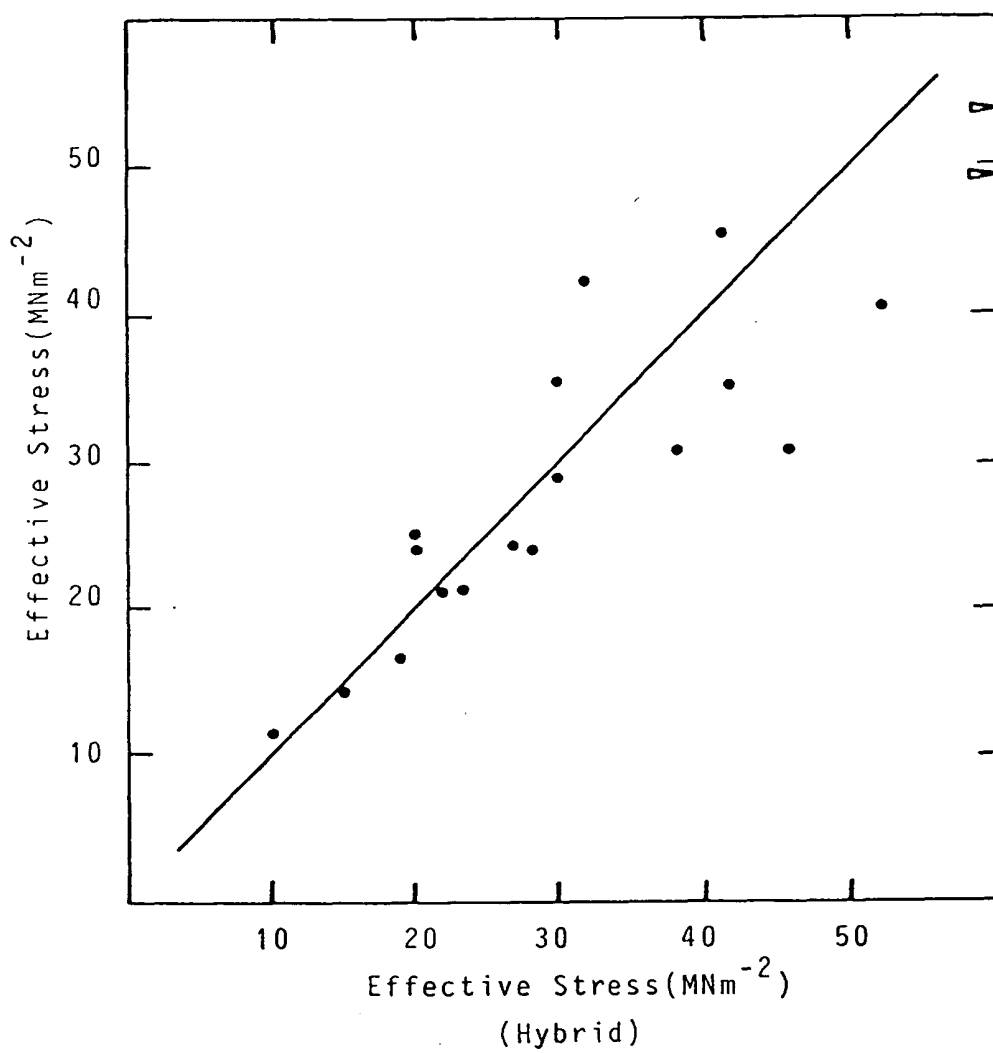


Figure 5.11. Experimental Flow Stress Vs. Predicted Flow Stress (Hybrid Model)

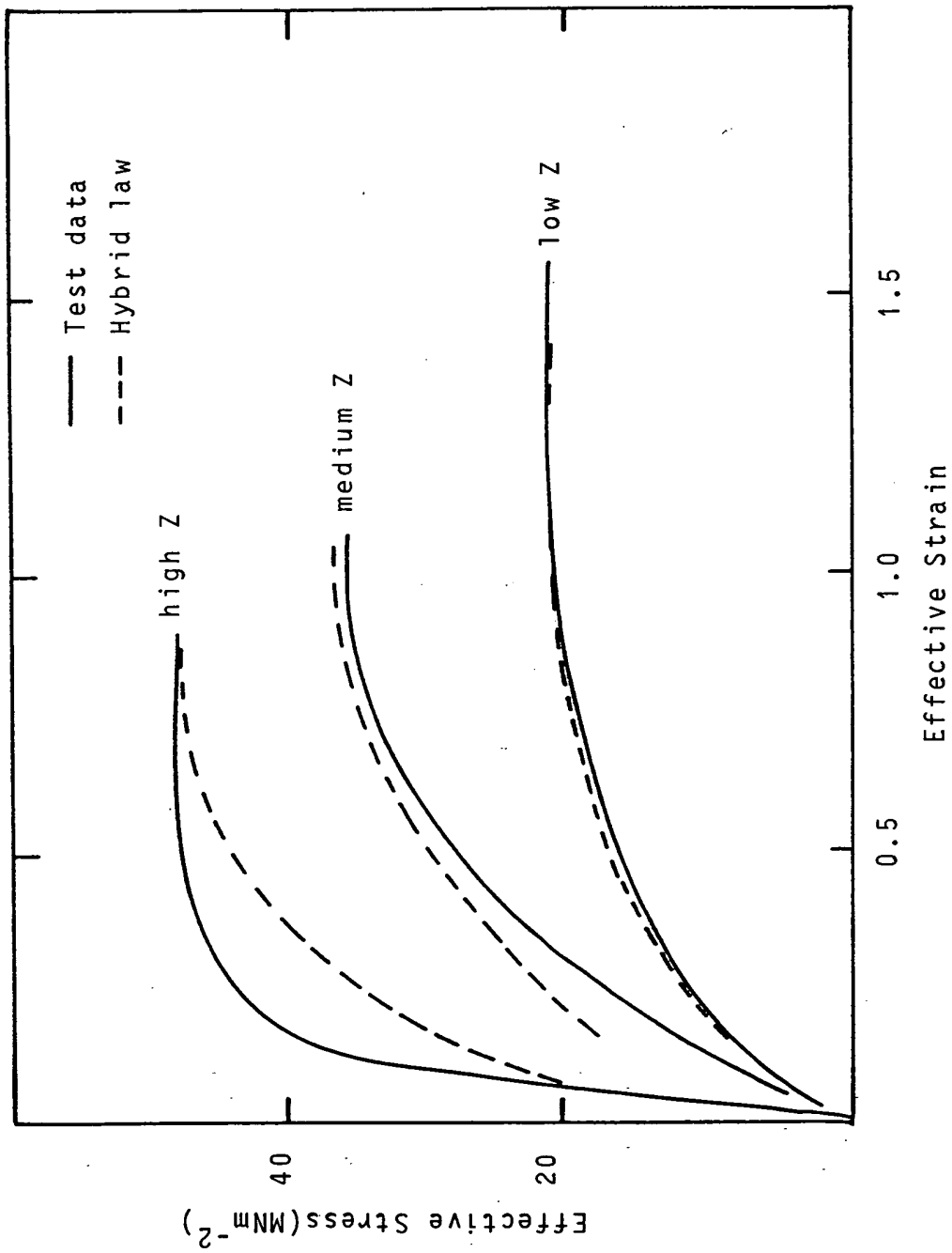


Figure 5.12. Experimental and Predicted Flow Stress Curves Using the Hybrid Model

This indicates that the use of one state parameter is insufficient to describe the evolution of structure in this alloy. The introduction of an additional structure parameter into the analysis was not possible at this stage. However, it is thought that if such an additional parameter was introduced and its evolution equation established, then the prediction range of the constitutive law would be extended.

5.2.4.4. Comparison of Constitutive Laws

The three constitutive laws discussed in the previous sections demonstrated the difficulty in correlating flow stress data over a wide range of temperature and strain-rate, especially when strain dependence was sought. This is reflected in the steady state flow stress results, listed in Table (5.8.), for each of the constitutive laws.

At low values of 'Z' each model provides predicted flow stress values in good agreement with the experimental data. However, as the value of 'Z' increases the predictive quality of the laws decrease. A clear comparison of the models is made in Figure (5.13.) where the calculated flow stress, using each equation, is plotted against $\ln Z$. At low 'Z' values the curves are almost coincident whilst at high 'Z' conditions, the curves diverge. Figure (5.13.) also illustrates the failure of the Anand model at high values of 'Z' as the curve moves away from the experimental data points. The Zenner-Hollomon and Hybrid models on the other hand retain reasonable correlation with their representative curves enveloping the experimental results. The

Table 5.8

Temperature (°C)	Strain-Rate (s ⁻¹)	Flow Stress (MNm ⁻²)	Flow Stress MNm ⁻²		
			Zenner	Anand	Hybrid
350	0.2098	45.60	33.88	48.69	40.98
400	0.2211	35.60	26.25	28.70	28.40
450	0.2080	25.10	20.15	17.70	20.20
500	0.2295	13.90	16.07	12.08	15.30
400	0.0409	23.90	20.52	18.15	20.70
350	0.0470	42.30	28.02	34.40	31.10
400	1.9052	35.60	34.61	52.54	42.30
350	1.9690	52.40	43.34	89.44	61.30
450	1.9007	28.00	27.73	32.24	30.61
500	1.9941	21.40	22.52	21.70	23.20
350	6.5400	53.20	48.64	123.90	75.90
350	19.5000	49.40	53.54	166.70	92.00
400	19.5000	39.50	44.50	96.87	64.20
400	6.2929	41.70	39.60	71.25	52.50
450	6.4004	31.50	32.43	44.83	38.30
450	19.5616	31.80	37.01	60.72	46.90
500	6.3846	24.40	26.59	29.81	28.90
550	19.3300	26.70	25.76	28.16	27.60
550	6.2974	20.90	21.90	20.77	22.40
550	2.0616	17.13	18.45	15.34	18.10
550	0.1469	13.00	11.89	7.48	10.80

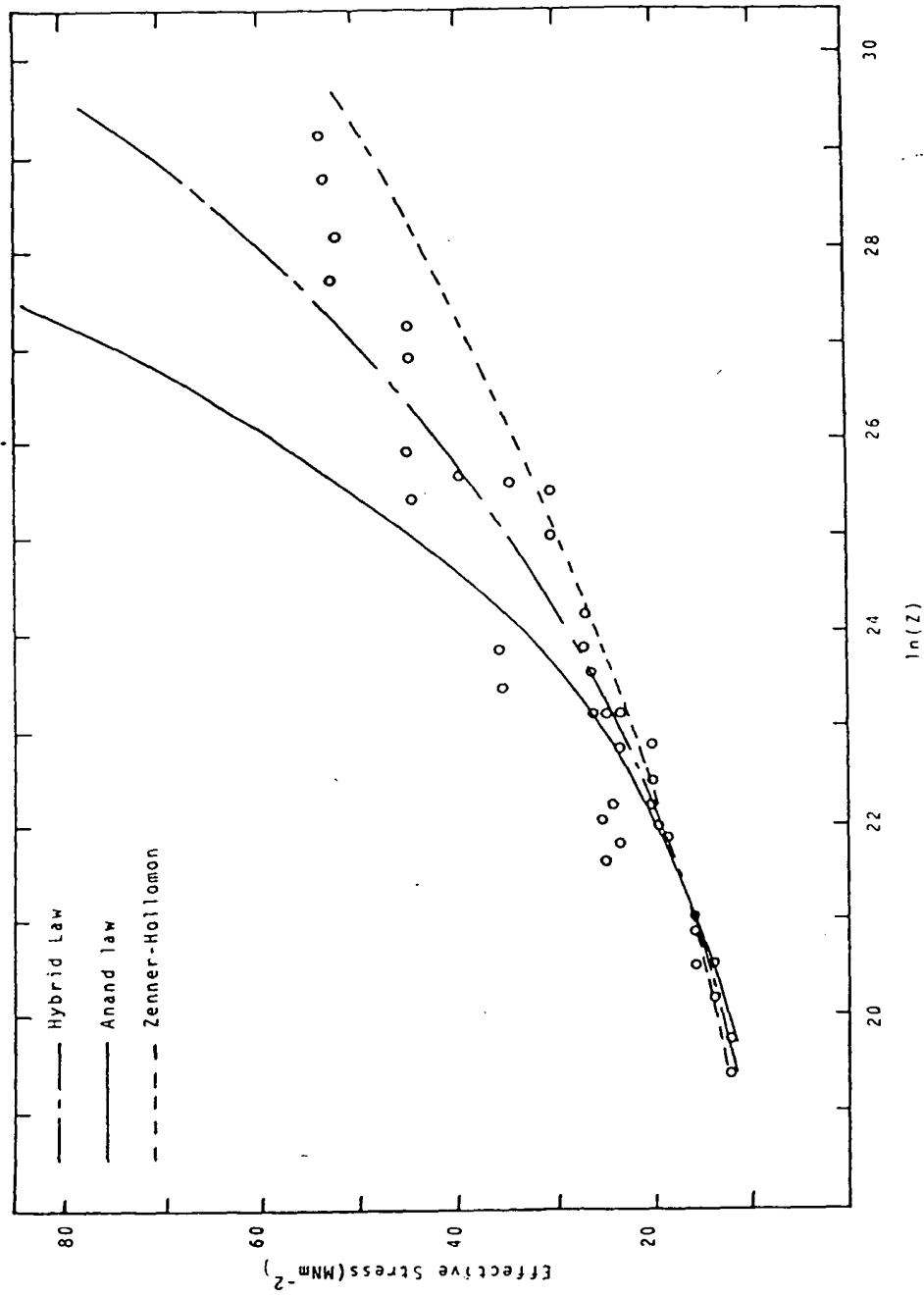


Figure 5.13. Comparison Between Zenner-Hollomon, Anand and Hybrid Models

Zenner-Hollomon equation provides an under estimate of the flow stress whilst the Hybrid model produces an over estimate.

The implication of the constitutive equation results is that, for steady state analysis, the Zenner-Hollomon equation remains superior and is, therefore, most suitable for adoption in the steady state extrusion analysis. Unfortunately, extrusion analysis using finite element modelling (FEM), for which the strain based constitutive laws were introduced, was not available for the present work. Therefore, assessment of these laws could not be made. However, it is thought that for successful results, using FEM, the strain based laws must be developed to include the minimum number of state variables and ultimately referred to a time base.

Conclusions

- (1) The developed torque during hot torsion testing of AA6063 increases with increase in strain-rate and decrease in temperature.
- (2) The calculated flow stress, using torque minimisation and effective radius analysis, shows dependence upon temperature and strain-rate, resulting from its integral relation to torque.
- (3) Hot torsion is a thermally activated process. The main softening mechanism operative during deformation is dynamic recovery.

- (4) Flow stress dependence upon temperature and strain-rate is most conveniently presented in the form of a constitutive equation. The most suitable equation being that of Zener-Hollomon.
- (5) A one parameter, mechanical equation of state, of the type proposed by Anand is not suitable for the heat treatable, AA6063, alloy.
- (6) A hybrid model, consisting of a Zener-Hollomon type kinetic equation and an evolution equation proposed by Anand possesses good predictive capabilities at low and medium 'Z' conditions.

5.3 Extrusion Data Analysis

The process conditions for extrusion of AA6063 were varied from 350°C to 550°C and exit speed from 30mm^s⁻¹ to 4800mm^s⁻¹. Extrusion ratios of 10:1 to 80:1 were utilised producing strain rates generally less than 50s⁻¹. This matrix of process condition was repeated for indirect extrusion and the direct extrusion of shapes.

The values of strain-rate and temperature rise obtaining from the extrusion trials were calculated using a computer program which has been developed by many workers^(7,9,11,14) and is outlined by Vierod⁽¹¹⁾ and Cooper⁽¹⁴⁾.

Observations made on the extrusion pressure revealed that it increased with ram speed and reduction ratio and with decrease in temperature. This observation is in keeping with the time and temperature requirements for dislocation movement which affects deformation movement by cross-slip and climb in the hot working regime.

5.3.1. Pressure-Displacement Curves

Examples of the pressure/displacement curves for both direct and indirect extrusion of AA6063 are shown schematically in Figure (5.14.). All of the curves show the characteristic rapid rise in pressure as the billet is upset to fill the container. During this stage a small amount of extrusion takes place, of the order of the extrudate

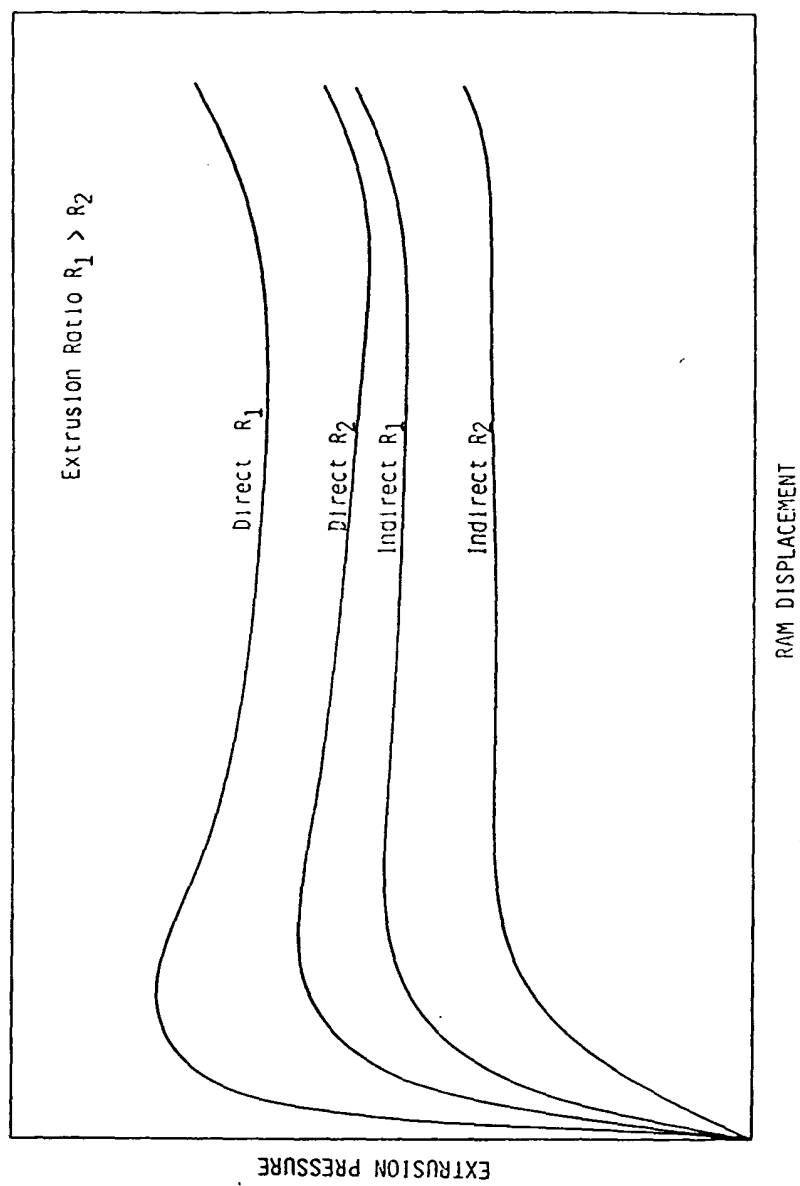


Figure 5.14. Load/Displacement Curves for Direct and Indirect Extrusion (Schematic)

diameter in length. Once the pressure/displacement locus reaches its peak, there is a gradual decrease in pressure throughout the extrusion cycle until near completion where a characteristic sharp increase in pressure occurs resulting from the excess pressure required, to affect radial flow, in the extrusion discard. It is usual practice to end the extrusion prior to this final pressure rise, as the formation of the deleterious funnel defect can occur during this time.

The gradual decrease in the pressure during quasi-static deformation results from the temperature rise generated during the extrusion cycle and is greater in those trials where high values of temperature compensated strain rate ('Z') are employed; that is, low temperature and high strain-rate.

The diminution of pressure during steady-state extrusion is further accentuated, in the direct mode of extrusion, with the decreasing length of the billet and the associated reduction in container wall/billet friction.

The prominent ' ΔP ' observed in some high strength alloys^(9,11,14) (Figure (5.15.)) was not generally detected in AA6063 extrusion, with the exception of processing at very high values of 'Z'. This is a consequence of the very high rate of dynamic recovery operative under normal hot working conditions in this alloy. The formation of ' ΔP ' at very high 'Z' processing is a result of the imposed storage of dislocations awaiting thermal activation to cross-slip and climb in order to accomplish annihilation⁽¹¹²⁾.

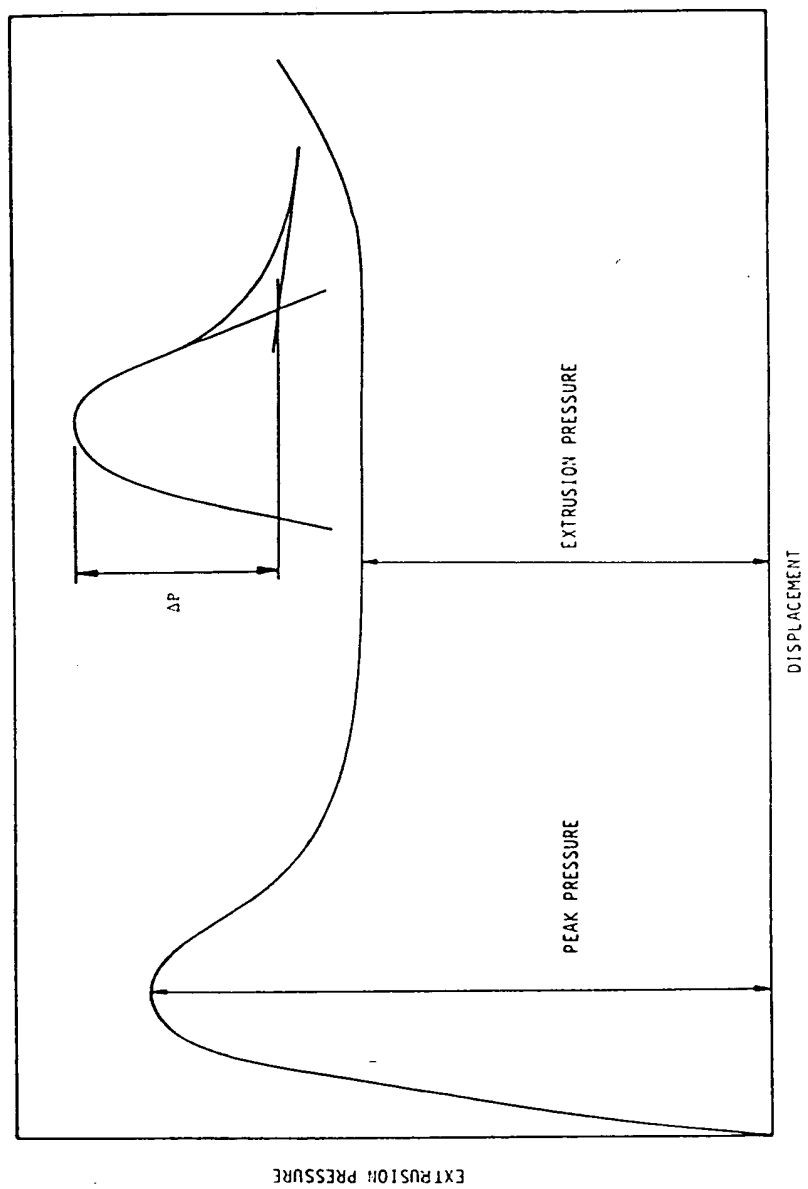


Figure 5.15. Schematic of ΔP in Load/Displacement Curves

Another feature which is evident from Figure (5.14.) is that the indirect mode of extrusion requires significantly less pressure to perform extrusion than the direct mode, at identical process conditions. This is due to the removal of the container wall/billet friction in the indirect mode. The effect of the larger deformation zone in the direct mode of extrusion might also demand greater pressure to produce a larger number of excess dislocations to accommodate deformation.

The direct extrusion of shaped sections was also observed to achieve a greater pressure than an equivalent reduction ratio, rod extrusion (Figure (5.16.)). This primarily results from the increased complexity in flow and a rise in the amount of redundant work performed.

5.3.2. Variation of Extrusion Pressure with Reduction Ratio

It was demonstrated in the previous section that extrusion pressure increased with reduction ratio. This dependence is plotted in Figure (5.17.a & b) at various temperatures for both modes of extrusion. A constant temperature has been assumed for each curve, which may not be strictly true. However, the work done to peak pressure and its conversion to heat does not have a profound effect, as it is relatively small and is balanced by heat flow to the tooling which is maintained at 50°C below the billet temperature. The results depicted in Figure (5.17.) show that the increase in pressure with log extrusion ratio at each temperature, is more or less linear and can be represented by an equation of the form:-

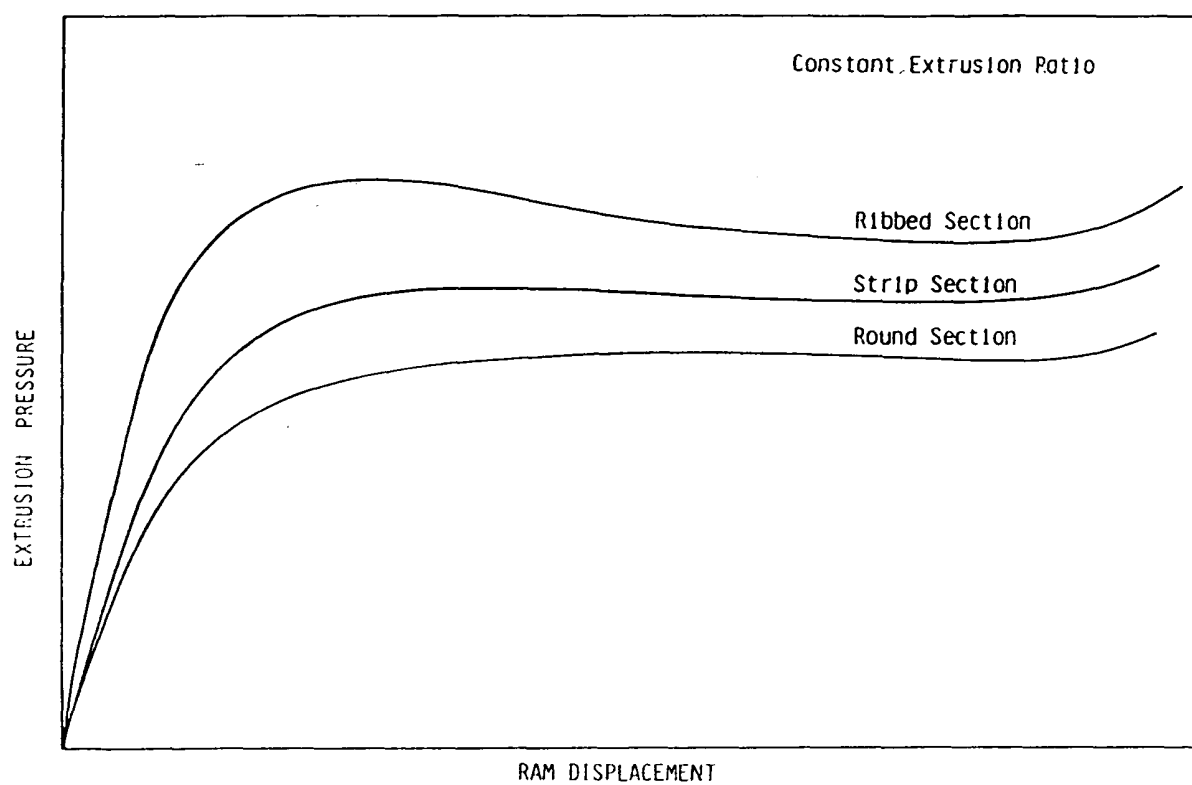


Figure 5.16. Load Displacement Curves for 60:1 Extrusion of Shapes (Schematic)

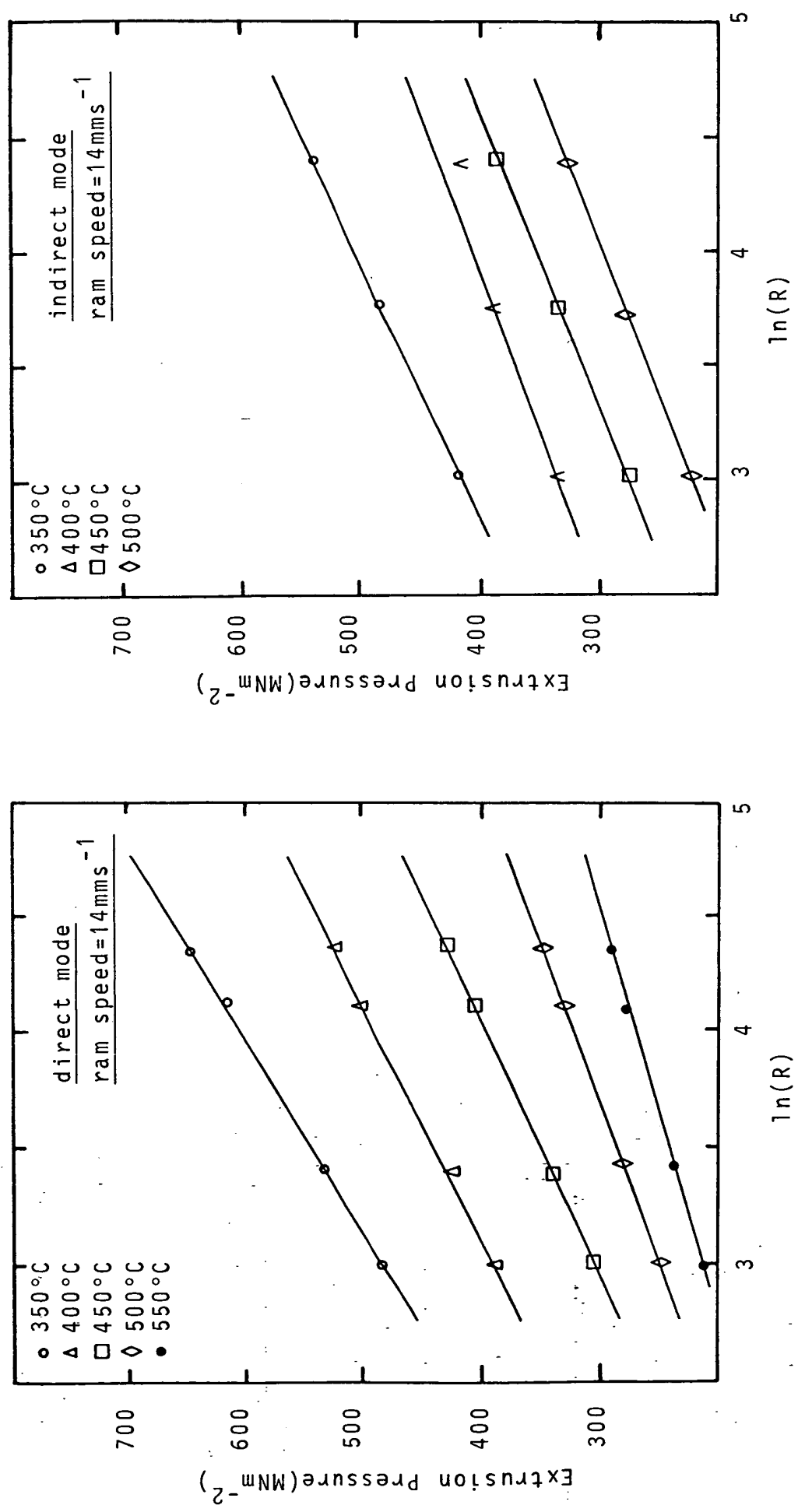


Figure 5.17. Variation in Extrusion Pressure with $\ln R$ (Direct and Indirect Mode)

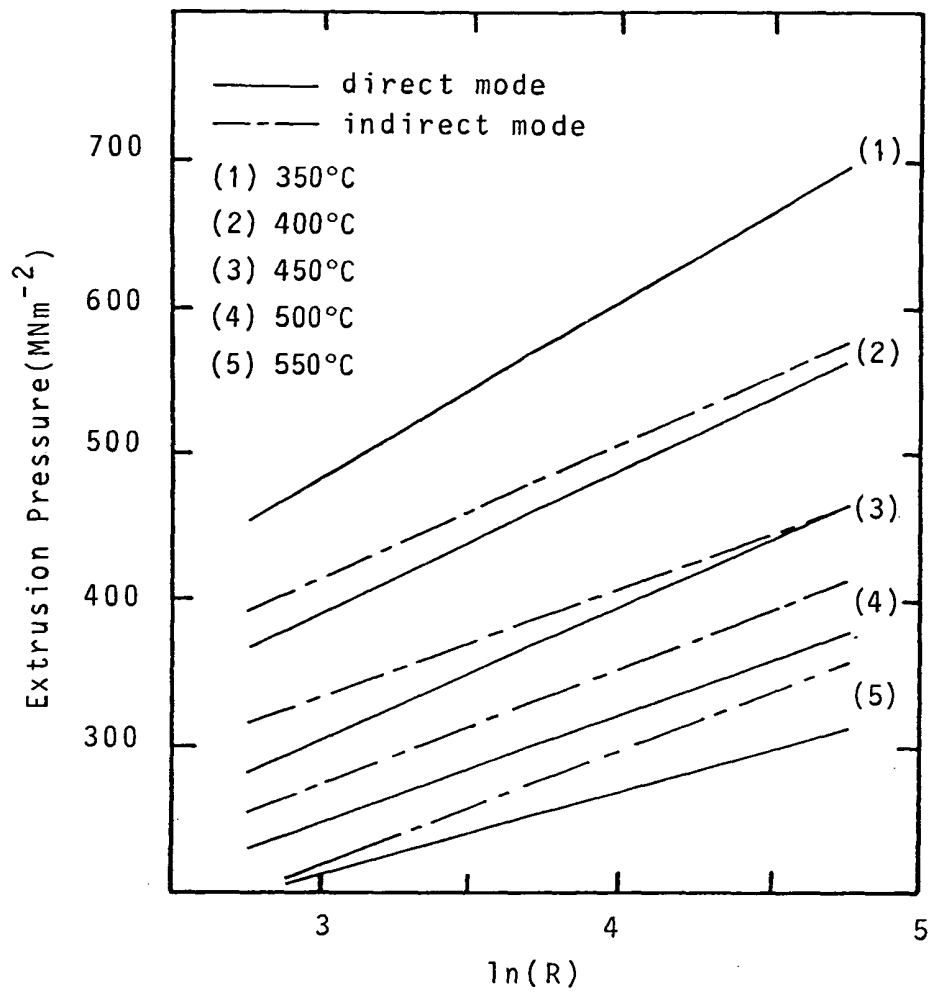


Figure 5.17(c) Variation in Extrusion Pressure with $\ln R$ (Direct and Indirect Mode)

$$P = A + B \ln(R) \quad -(5.11.)$$

This relationship has been applied to aluminium extrusion data by previous workers(7-9,11,13-15) for heat treatable and non-heat treatable alloys demonstrating high correlation. The extrusion data from the present study of AA6063 was also fitted to Equation (5.11.), the results of which are given in Table (5.9.).

Table 5.9. Extrusion Constants

P=A+B. lnR	350	400	450	500	550
Direct A	120.0	100.0	37.5	32.5	42.5
B	121.0	98.0	90.0	73.0	57.0
cc	.976	.980	.991	.990	.987
Indirect A	142.5	15.0	32.5	-22.5	-
B	91.5	94.0	80.0	80.0	-
cc	.968	.977	.984	.987	-

Again, a good fit to the experimental data was obtained which is demonstrated by the high values of the correlation coefficients. The value of 'B' in Equation (5.11.) is a measure of the pressure sensitivity to the magnitude of deformation and is higher for the direct extrusion mode. The reason for a lower pressure sensitivity in the indirect mode of extrusion originates from the more homogenous nature of flow during deformation and the lesser extent of redundant

work. In addition, the larger deformation zones present in direct extrusion require the generation of a greater number of dislocations to maintain the balance for the thermally-activated events obtaining during the extrusion process.

The constant 'A' in Equation (5.11.) increases markedly with temperature and so appears to represent the pressure dependence upon temperature. However, these values are found by extrapolation of the extrusion data and lie outside the known validity boundary $2.5 < \ln R < 5$. Thus a physical significance to these values can not safely be made.

In Figure (5.17.c) a comparison is made of the pressure/extrusion ratio relationship between the two modes of extrusion. It is clear that both modes possess similar dependence of pressure upon the magnitude of deformation. The difference in the extrusion pressure between direct and indirect extrusion reduces as the initial billet temperature is raised. This reduction in pressure difference between extrusion modes is a direct result of the increased thermal energy allowing the dislocations to cross-slip and climb more readily, reducing dislocation storage, and the redundant work prevalent in the direct extrusion mode.

This observation of increasing efficiency in the direct extrusion mode with increase in initial billet temperature agrees in principle with trends observed by McShane⁽⁴⁾ working with powders of a similar composition to AA6063.

5.3.3. Variation of Extrusion Pressure with Initial Billet Temperature

The dependence of extrusion pressure on the initial billet temperature at increasing reduction ratios is shown in Figure (5.18) for both modes of extrusion. The figure shows that for each reduction ratio the extrusion pressure decreases, with increasing temperature (as the flow stress decreases). The form of the curves indicates that a power relationship of the form:-

$$P = C T^D \quad \quad \quad -(5.12)$$

might exist, where 'C' & 'D' are constants at each extrusion ratio. A regressional analysis was performed on the data and the results are listed in Table (5.10).

Table 5.10. Extrusion Constants

$P = CT^D$	80:1	40:1	20:1
Direct C	47.511E6	41.548E6	34.980E6
D	-1.9059	-1.9059	-1.9059
cc	.946	.956	.960
Indirect C	1.975E6	3.706E6	13.247E6
D	-1.3985	-1.5265	-1.7647
cc	.962	.971	.969

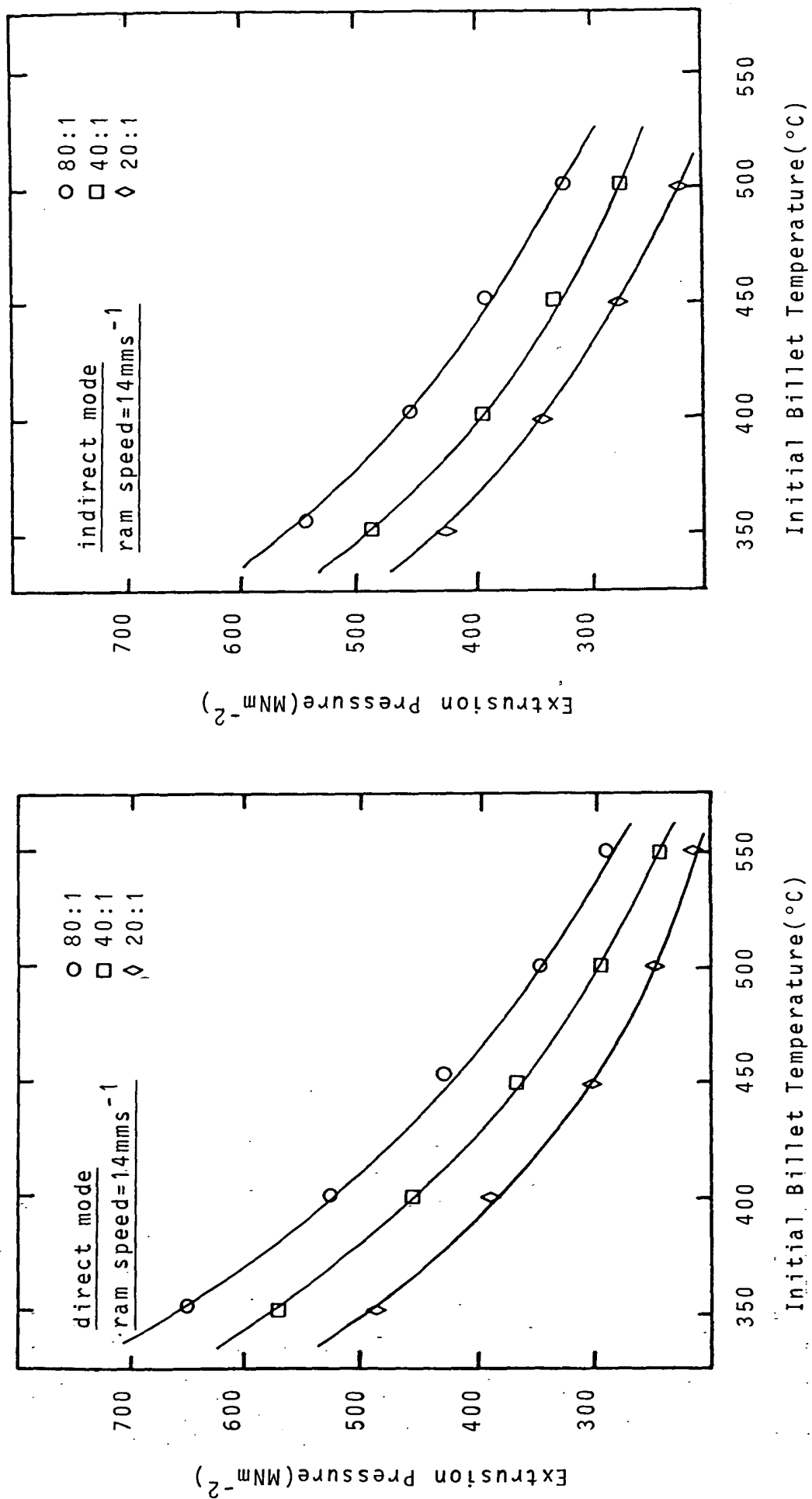


Figure 5.18. Variation in Extrusion Pressure with Temperature (Direct and Indirect Mode)

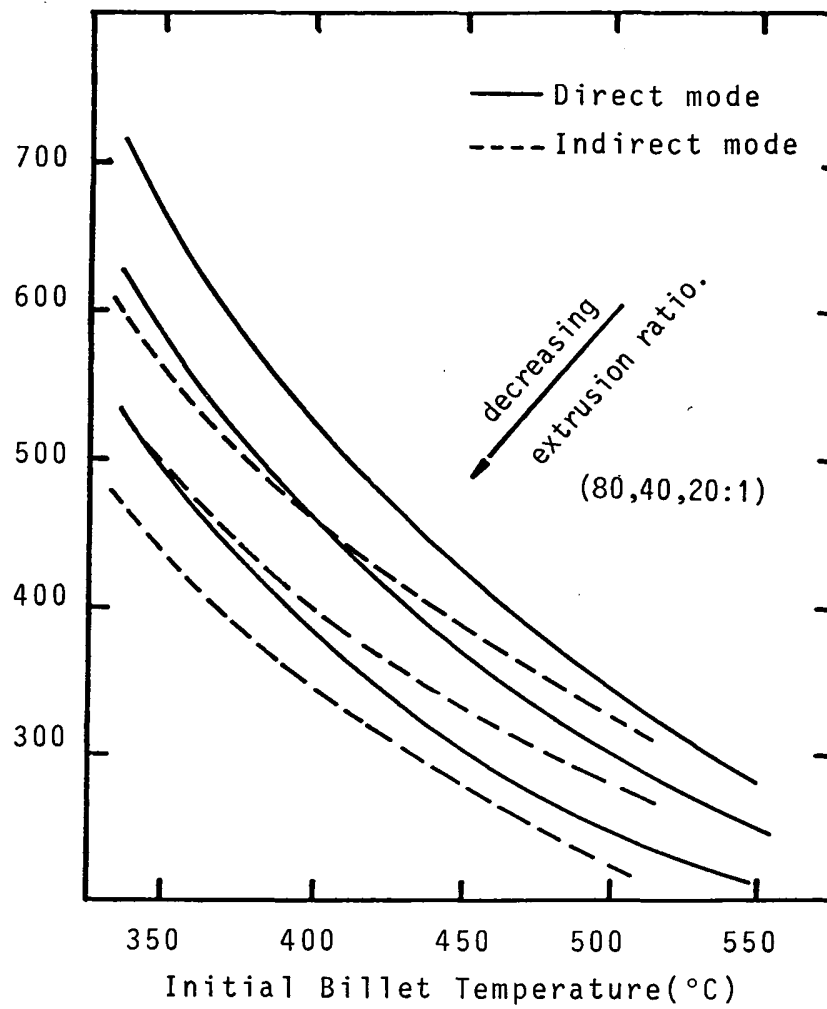


Figure 5.18(c) Variation in Extrusion Temperature (Direct and Indirect Mode)

Previous workers have reported linear⁽⁷⁾, reciprocal⁽⁵⁾ and exponential⁽¹⁾ relationships to fit the extrusion pressure/temperature data. However, the extrusion conditions under which these trials were performed were not necessarily similar to the present work under discussion. The use of Equation (5.12) was also favoured by Vierod⁽¹¹⁾ who pointed out that previous workers^(5,7) had employed a constant container temperature and so precise temperature data was less likely to be obtained. The current practice was to maintain the container temperature at 50°C below that of the billet, thus producing a more uniform heat balance at each extrusion temperature⁽¹¹⁾.

Equations such as (5.11) and (5.12) are of limited use and prove to be cumbersome in practice. To use these empirical equations, numerous matrices of constants are required to cover the range of possible working conditions. In addition, to obtain the constants, extensive extrusion trials must be performed, which may not be feasible industrially when developing new alloys.

The use of the torsion test to model the extrusion process, which was discussed earlier, conveniently permits the extrusion process to be defined through fewer equations. This is achieved by introducing the temperature compensated strain-rate (Z) into the extrusion analysis.

5.3.4. Variation of Extrusion Pressure with $\ln Z_p$

The assumption that extrusion is a thermally activated, hot working process similar to torsion testing is to some extent confirmed by the many excellent correlations obtained between the peak pressure and $\ln Z_p$ (7-9,11,14). Since $\ln Z_p$ effectively combines both the temperature and strain rate parameters, it has been found that a single line relationship can be obtained over a range of temperatures and strain-rates for a constant reduction ratio. The results from the present study on AA6063 are shown plotted in Figure (5.19). The results from these graphs are listed in Table (5.11).

Table 5.11. Extrusion Constants

$P = A + B \cdot \ln Z_p$	A	B	CC
Ratio - Mode			
80:1 - Direct	-936	50.5	0.989
20:1 - Direct	-721	40.5	0.981
80:1 - Inirect	-650	37.5	0.979
20:1 - Indirect	-699	37.0	0.988

The high correlation coefficients indicate that the relationship is applicable to all temperatures and strain-rates considered. The lines drawn through the correlated data show that a greater pressure is

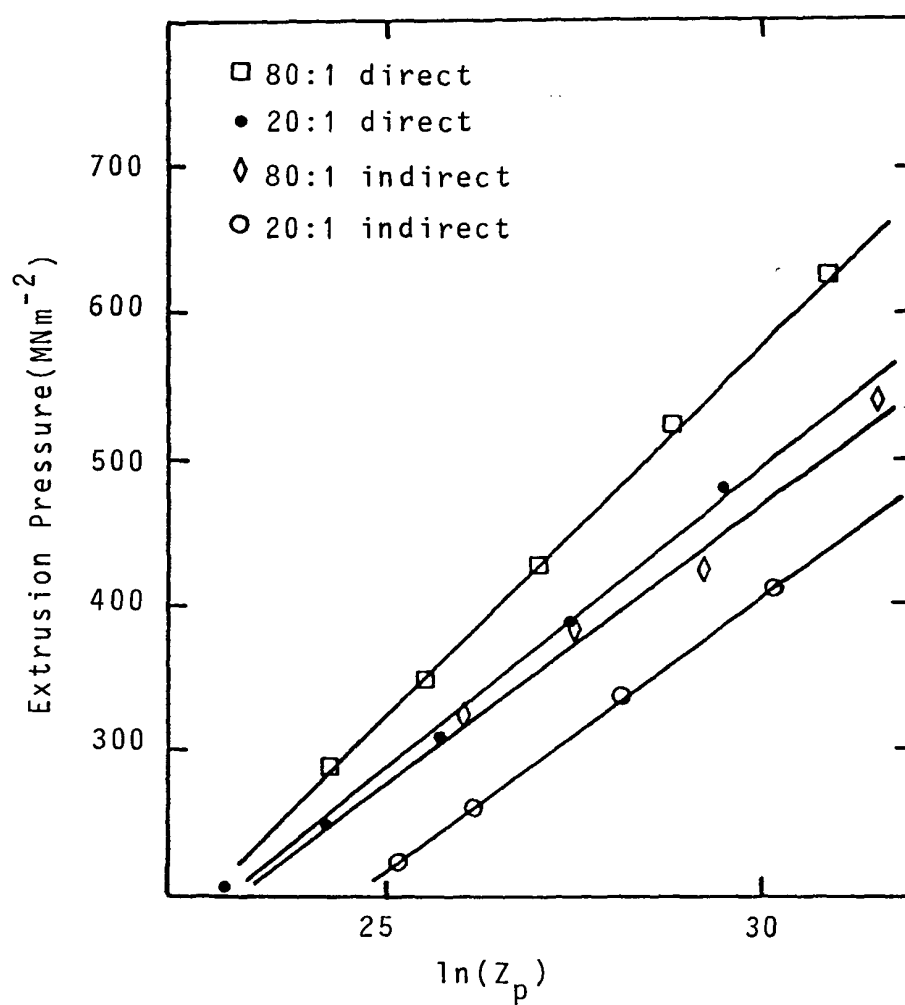


Figure 5.19. Variation in Extrusion Pressure with $\ln Z_p$

required to extrude at higher reduction ratios and, at an equivalent magnitude of reduction, the pressure required is less when processing in the indirect mode. This form of equation thus possesses the required characteristics to define the extrusion pressure required to affect deformation. Vierod⁽¹¹⁾ working with alloys in the AA2XXX series found that by modifying the equation:-

$$P = A + B \cdot \ln Z_p \quad -(5.13)$$

plotted in Figure (5.19), to include the hot working constant 'A', the reformed equation possessed alloy independence. The modified equation:-

$$P = A' + B' \cdot \ln \left(\frac{Z_p}{A} \right) \quad -(5.14)$$

is applied to the extrusion data for 20:1 (direct mode) using AA6063 and is plotted together with Vierod's AA2XXX series results in Figure (5.20). From this figure it is evident that there is considerable disparity between the results. Putting the inevitable differences in the analytical processing of the extrusion data aside, it is clear that the peak pressure's sensitivity to $\ln (Z_p/A)$ is the same (i.e. alloy independent) for both sets of results, demonstrated by the equivalent gradients of the lines. The transposition of the AA2XXX series curve to a higher magnitude of pressure, results from the alloy-dependent break-through pressure. Sheppard⁽¹¹⁶⁾ reporting on the extrusion of complex aluminium alloys pointed out that as much as 40% excess pressure may be required over that of steady state pressure to initiate extrusion. The excess pressure required for the softer alloys such as AA1XXX series and

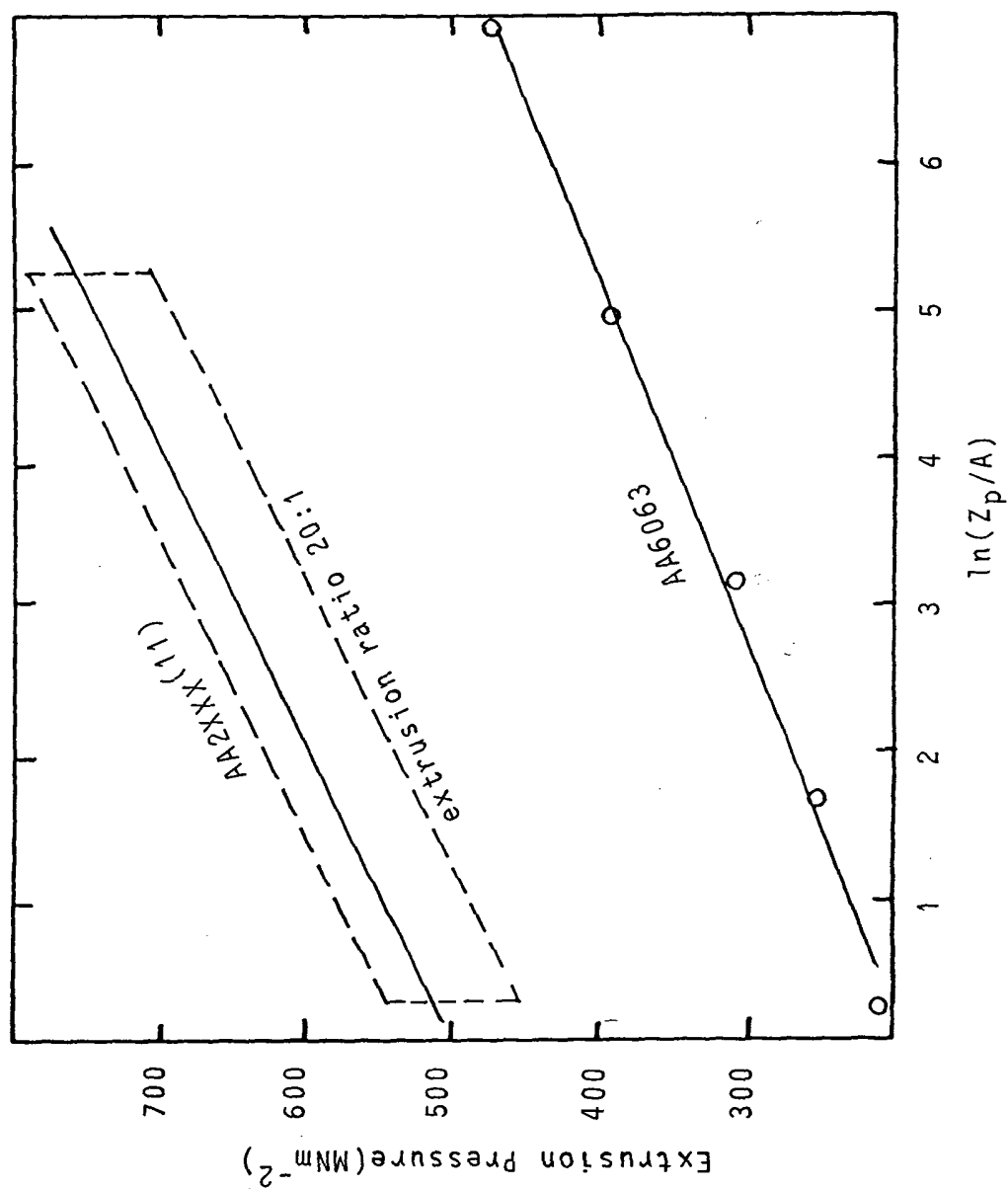


Figure 5.20. Comparison of Extrusion Pressures in AA6063 and AA2014 with Respect to $\ln(Z_p/A)$

AA6063 is negligible under normal processing conditions.

In addition to the departure of results derived from requirements of breakthrough pressure, it is unlikely that the generation and recovery mechanisms operative - which dominate dislocation activity and hence deformation - are adequately represented in $\ln (Z_p/A)$ during transition to steady state. This is because the hot working constants are defined for (and only valid at) steady state deformation.

Equations (5.13) and (5.14) are of considerable use but still require numerous extrusion trials to define the equation constants for each extrusion ratio, and so again may prove tedious to use practically.

5.3.5. The General Pressure Equation

The equations applied to the extrusion data in the previous sections have shown that the total extrusion pressure is dependent on the strain rate, extrusion ratio, and extrusion temperature. Mathematical models based on slip line field theory and upper bound solutions predict⁽¹¹⁷⁾ that a general pressure equation relating to the above extrusion parameters will be in the form:-

$$P/\sigma = a + b \cdot \ln (R) \quad \text{---(5.15)}$$

where:

P = Extrusion Pressure (MNm^{-2})
 σ = Mean Equivalent Flow Stress Rate (MNm^{-2})
 R = Extrusion Ratio
 and a, b = Extrusion Constants.

Assuming that the value of σ is equivalent to the steady state flow

stress defined by the hot working equation:-

$$\sigma = \frac{1}{\alpha} \ln \left(\frac{Z^{n-1}}{A} + \left(\frac{Z^{2n-1}}{A} + 1 \right)^{0.5} \right) \quad -(5.16)$$

and since the two terms within the bracket are approximately equal, the equation is reduced to the more simple form:-

$$\sigma \approx \frac{0.7}{\alpha} + \frac{1}{\alpha n} \ln \frac{Z}{A} \quad -(5.17)$$

Equation (5.17) can now be substituted into Equation (5.15) to form the extrusion pressure equation:-

$$P = \left(\frac{0.7}{\alpha} + \frac{1}{\alpha n} \ln \frac{Z}{A} \right) (a + b \ln(R)) \quad -(5.18)$$

The equation has embodied in it dependence upon temperature, strain-rate and extrusion ratio and so defines extrusion pressure under a given set of conditions through one equation. The similarity between this equation and the empirical equations described elsewhere⁽¹¹⁸⁾ indicates that extrusion pressure may be more conveniently described by an equation of the form:-

$$P = \frac{1}{\alpha n} \left(A' + B' \ln(R) + C' \ln \frac{Z}{A} \right) \quad -(5.19)$$

where A', B' & C' are extrusion constants.

The validity of this form of equation to describe extrusion pressure as a function of temperature, strain-rate and extrusion ratio was assessed using multiple regression analysis on extrusion data collected for AA6063. The results from this treatment are plotted in

Figures (5.21-22) and the constants determined for each extrusion mode are listed in Table (5.12) together with the correlation coefficients.

Table 5.12. General Pressure Equation Constants

Mode	A'	B'	C'	cc
Direct	22.65	4.66	10.530	0.976
Indirect	15.00	4.66	8.643	0.962

Hot working constants used:

ΔH	=	141550
α	=	0.04
n	=	5.385
$\ln A$	=	22.5

The high correlation for each extrusion mode indicates that the derived equation is applicable over the range of extrusion conditions considered. This is clearly demonstrated in Figures (5.21-22) which shows the experimental pressure against the predicted pressure. The use of only three extrusion constants in Equation (5.19) makes the interpretation of any physical significance to the values difficult, as each constant may include the effects of more than one deformation controlling mechanism. The global minimisation of errors in the regression analysis might also mask the true meaning. However, with this in mind, it is clear that each constant can be regarded as governing the contributions of each variable to the total work of deformation. The term $B' \cdot \ln R$ is proportional to the useful work, since $\ln R$ is simply a measure of the logarithmic (true) strain suffered by the material and A' is a measure of the redundant work, which can be seen, is higher for the direct mode of extrusion. The

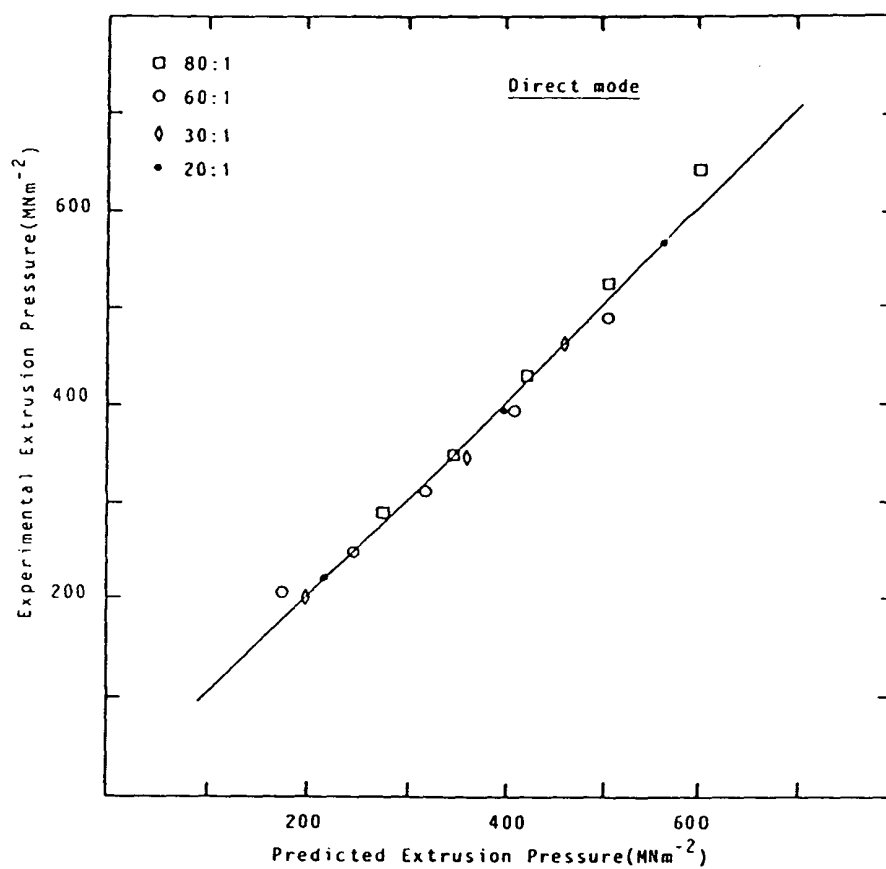


Figure 5.21. Experimental Extrusion Pressure Vs Predicted Pressure from General Pressure Equation (Direct)

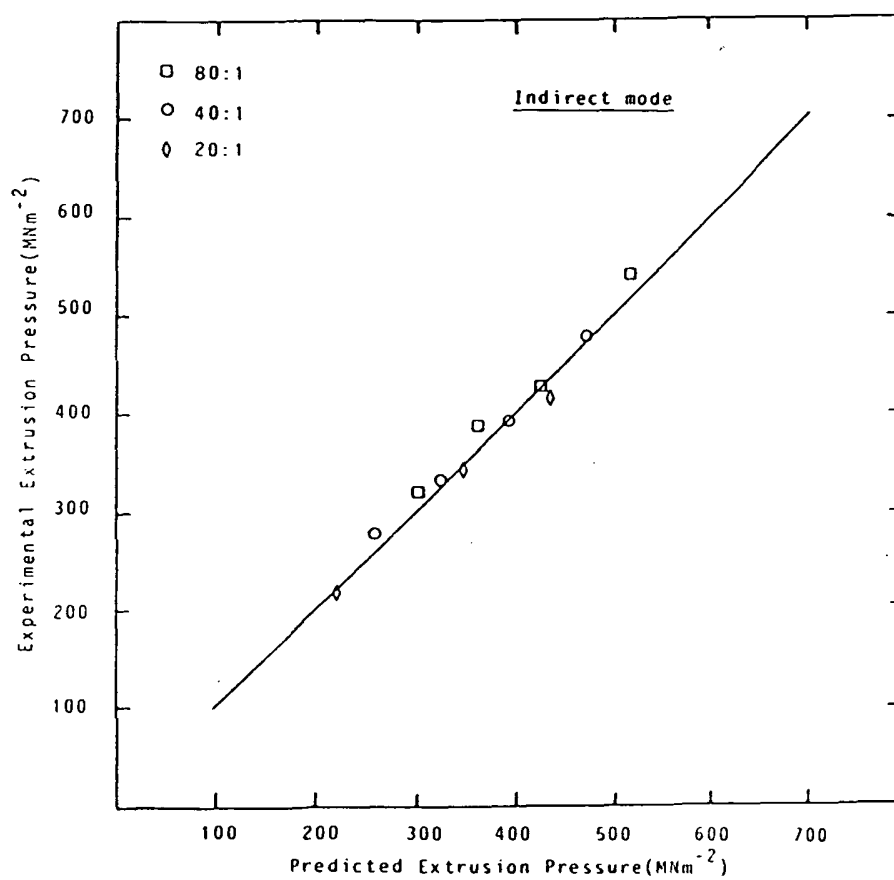


Figure 5.22. Experimental Extrusion Pressure Vs Predicted Pressure from General Pressure Equation (Indirect)

final term $C' \ln(Z/A)$ introduces the hot working constants into the equation which effectively allows the material properties for flow stress characteristics to be included.

The principal limitations of Equation (5.19) are that it applies to billets of specific dimensions and does not explicitly include a frictional term. These features can be included⁽¹¹⁾ but proved difficult in the present work for the following reasons. Firstly, the introduction of billet geometry into the equation was not possible because constant billet dimensions (75mm ϕ by 100mm) were used throughout the research programme.

The introduction of a friction term was not possible due to the difficulties in determining an accurate value for the coefficient of friction from the analysis of pressure decrease with billet length, determined from the quasi-static portion of the pressure displacement curve. This is because of the negligible change in the pressure during this time, for AA6063.

The former limitation of Equation (5.19) is not of any significant concern as the billet geometry would generally be maintained constant by an extruder. However, the omission of an explicit term for friction does give cause for concern, but was considered better than including a 'misleading' value for the coefficient of friction. Of course this contribution is included in the A' term for redundant work.

5.4. The Effect of Section Shape on the Pressure in Direct Extrusion

It has been reported that^(5,86,87) an additional increment of pressure is required, at a constant extrusion ratio, to extrude a section of complex design over that of equivalent circular cross section. During the current research, shaped sections were extruded and this phenomenon was observed. In Figure (5.16) the pressure displacement curves are shown for rod, strip and ribbed sections (R=60:1, billet temperature 450°C), clearly demonstrating this phenomenon. This increment of pressure is required over the range of temperatures considered (Figure (5.23.b)). When the peak pressures are plotted against $\ln Z$ a linear relationship of the form:-

$$P = A + B \ln Z_p \quad \text{---(5.13)}$$

is apparent.

Performing a regression analysis on the AA6063 extrusion data indicated a good correlation of the results. These results are listed in Table (5.13). Over the range of $\ln Z$ considered, the functions relating 'P' and ' $\ln Z$ ' for each section are more or less co-linear (Figure (5.23.a)).

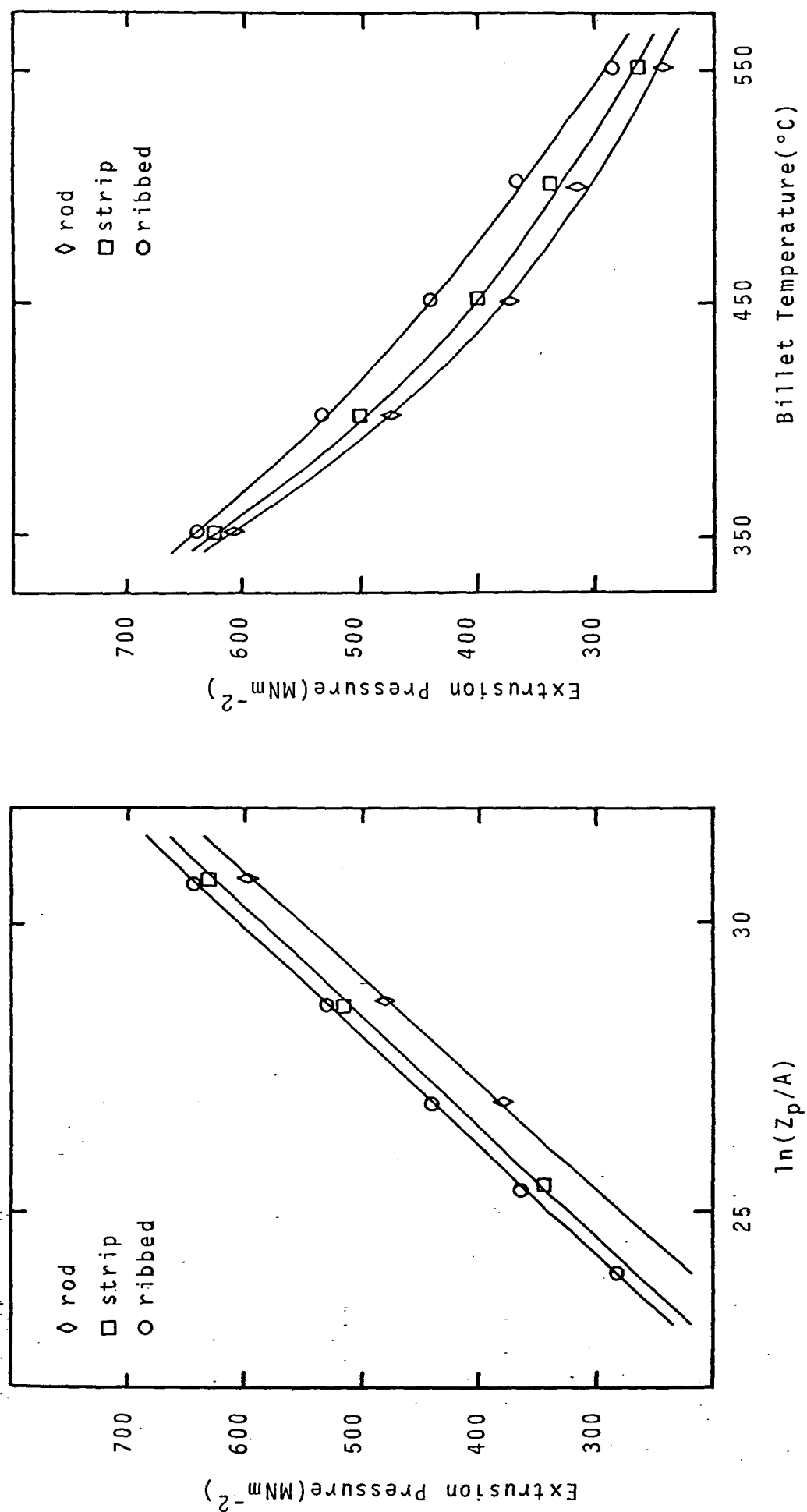
Figure 5.23. Effect of Section Shape on Extrusion of Pressure with Respect to $\ln Z_p/A$ and Temperature

Table 5.13. Extrusion Constants (Shapes)

$P = A + B \cdot \ln Z$	A	B	cc
Shape			
Round	-1100	55	0.988
Strip	-1070	55	0.991
Ribbed	-1050	55	0.995

The extrapolated value for the extrusion constant 'A' shows an increase with complexity of the section. This suggests that the amount of redundant work increases in shaped extrusion. However, Sheppard and Wood⁽⁸⁷⁾ pointed out that the value of 'Z', which is calculated from averaged process parameters, must vary with section complexity due to local distortions which occur at the die entry producing large variations in 'Z'. It would therefore be useful to 'correct' the value of 'Z' to allow for the required pressure increment.

One method utilised to compensate for the effect of differing shape has been the concept of peripheral ratio (Γ). This parameter is defined as: the ratio of the extrudate perimeter ' p_e ' to the circumference of a rod ' p_r ' having the same reduction ratio. The peripheral ratio, together with the section geometry of the dies used in the present study, are given in Figure (5.24). Wood⁽⁵⁾ used the

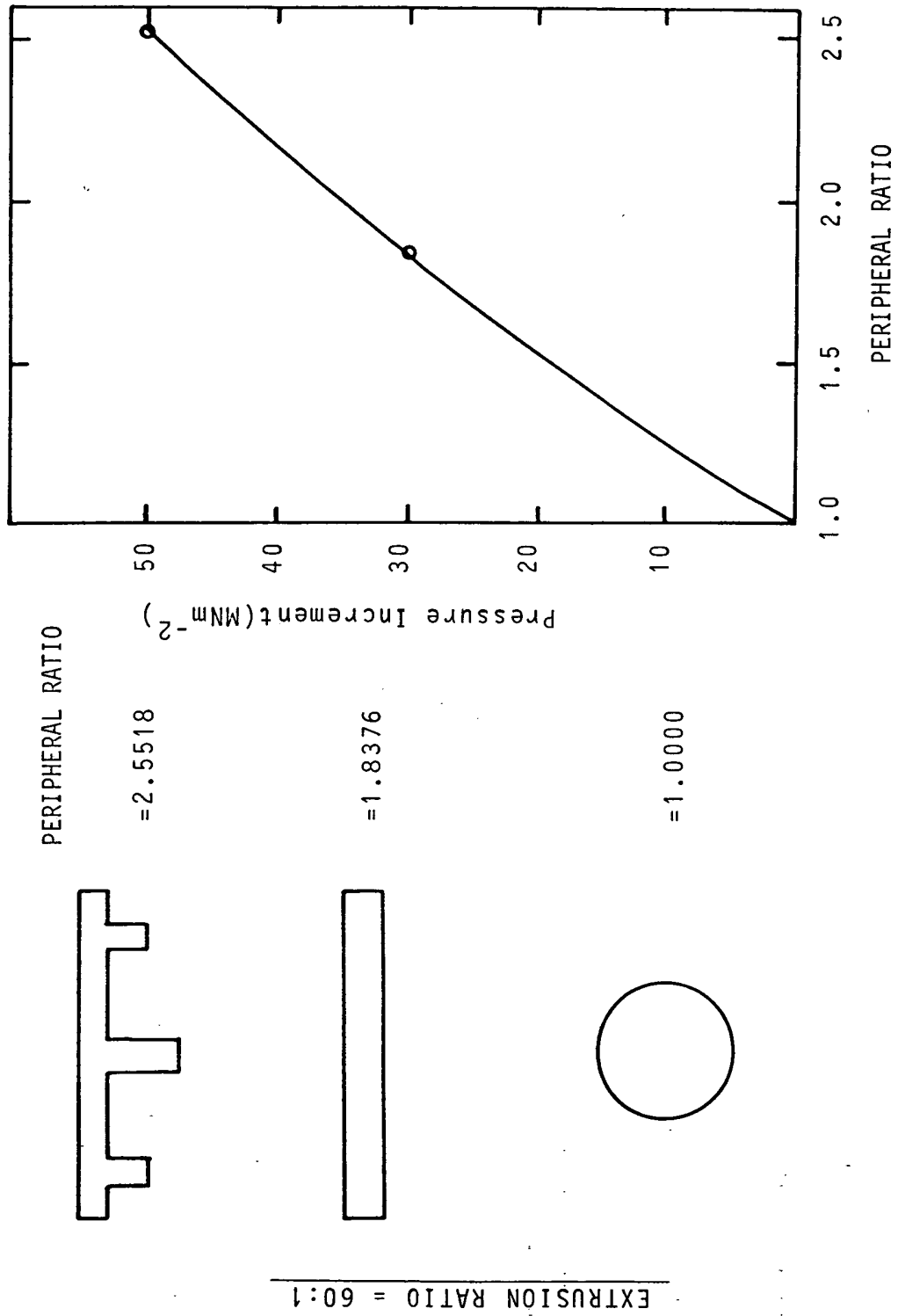


Figure 5.24. Schematic Sections Extruded and Corresponding Peripheral Ratio

peripheral ratio to obtain a modified extrusion ratio(R') defined by:-

$$R' = \Omega R \quad \text{---(5.20)}$$

where: $\Omega = (p_e/p_r)^{0.5} = \Gamma^{0.5}$
 R = extrusion ratio.

Wood then used this to obtain a correction factor to modify the temperature compensated strain-rate (Z) through an equation of the form:-

$$\ln R' = a \cdot \ln Z + b$$

where: a, b are constants
 $\ln Z$ is the shape correction factor.

In the present work it was found that the best correlation of data was obtained when the peripheral ratio (Γ) was applied directly to Equation (5.13) resulting in a modified form given by:-

$$P = A + B \ln(\Gamma Z) \quad \text{---(5.21.a)}$$

In this equation the extrusion constants A and B are derived from rod extrusion results (i.e. $\Gamma=1$).

The results of predicted pressure using this equation can be seen in Figure (5.25) together with experimental data, clearly demonstrating that this form of equation produces an excellent prediction of the

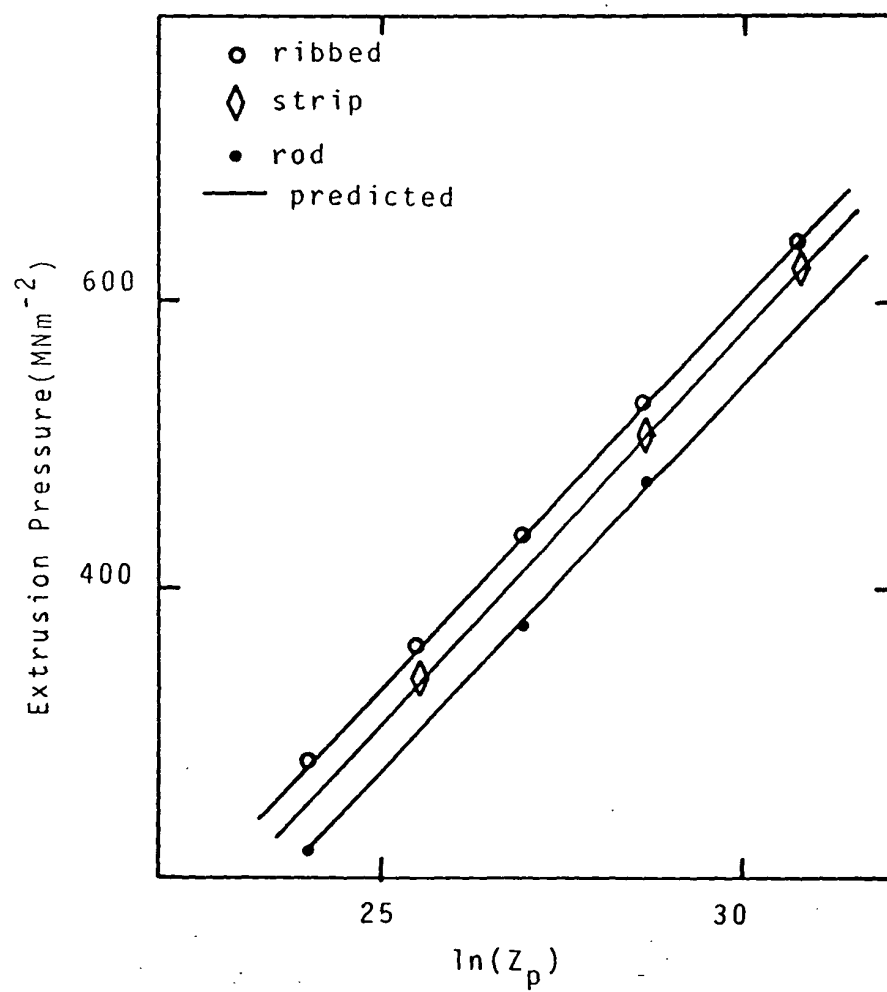


Figure 5.25. Predicted Extrusion Pressures for Shaped Extrusion.

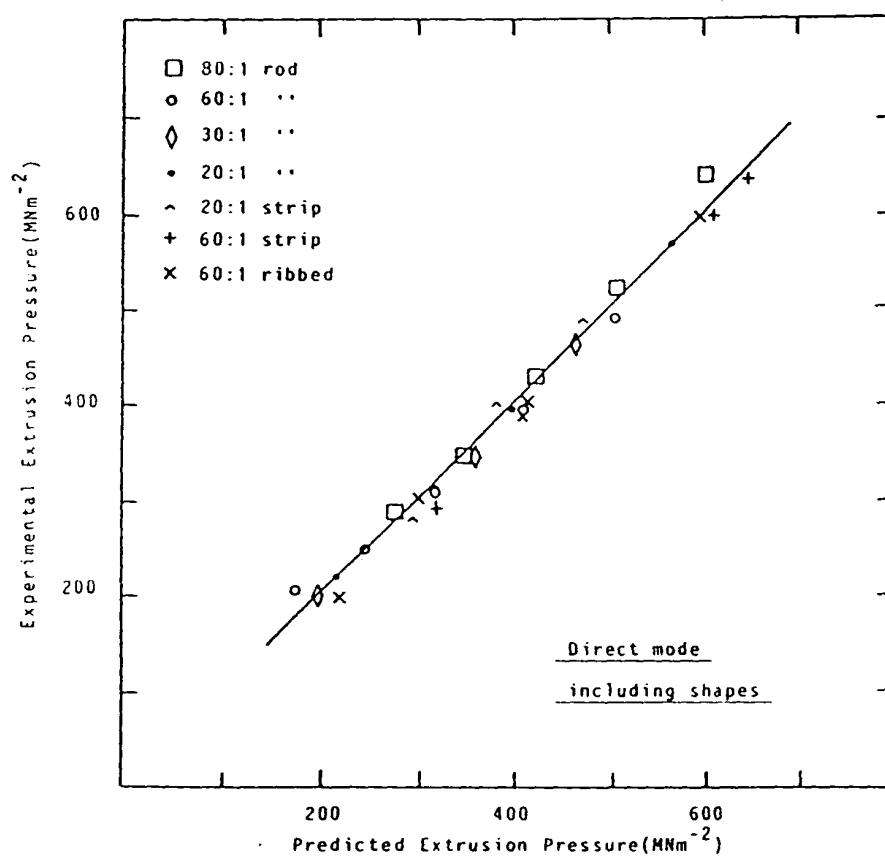


Figure 5.26. Predicted Extrusion Pressures for Shaped Extrusion Using a Modified General Pressure Equation

required extrusion pressure for rod and shaped products.

The use of peripheral ratio (Γ) can be extended by including it in the general pressure equation derived in Section 5.3.5. This results in the following modified equation:-

$$P = \frac{1}{\alpha n} \left(A' + B' \ln R + C' \ln \frac{\Gamma Z}{A} \right) \quad -(5.22)$$

where the constants are found for rod extrusion ($\Gamma=1$) as described in Section 5.3.5.

In Figure (5.26) the experimental pressure is plotted against the predicted pressure, for rod, and shaped extrusions, using Equation (5.22). Again, a good correlation exists between the results.

It is evident that the use of peripheral ratio to modify 'Z' in shaped extrusion is simple and effective. However, it requires further extensive research to confirm these findings by including a wider range of peripheral ratio and constant peripheral ratio with varying design. It is also envisaged that this treatment might be extended to include multi-hole dies.

Conclusions

- (1) Extrusion pressure increases with decrease in temperature and increase in reduction ratio, and is greater when extruding in the direct mode.

- (2) Extrusion is a thermally activated deformation process similar to torsion, at hot working temperatures. This validates the use of the torsion test to model extrusion.
- (3) A general pressure equation incorporating the hot working constants, and therefore flow stress characteristics, can be established
- (4) An increment of pressure is required to extrude a section of complex design over that of equivalent circular cross section.
- (5) The effect of section shape can be incorporated in the general pressure equation by correcting the 'Z' value using the peripheral ratio.

CHAPTER SIX

CHAPTER SIX

Material Flow and Structural Development During Extrusion

6.1. Structural Investigation

The evolution of structure and sub-structure at all stages of processing, from as-cast to the final-T1 extrusion product, has been investigated for the aluminium alloy AA6063. Also presented are the room temperature tensile properties of the extrudates in the -O, -T1, T4 and T6 conditions, produced at varying process conditions.

6.1.1. Introduction to AA6063

The Aluminium Association's chemical composition limits of AA6063 have a broad range for alloy and impurity elements, the major alloying elements being magnesium (Mg) and silicon (Si) with limits of 0.45-0.90% and 0.2-0.6% respectively. These elements combine forming the stoichiometric compound magnesium silicide (Mg_2Si). It is the Mg_2Si that makes the AA6XXX series alloys heat-treatable, capable of reaching full-T6 properties. Of the remaining elements making up AA6063, the most important is iron (Fe) which is present in all AA6XXX series alloys as a controlled impurity. Silicon will combine with iron, to form an AlFeSi phase, in preference to magnesium, thus creating problems in both final material properties and the alloy's surface generation.

A more detailed examination of AA6XXX series alloys with special attention to AA6063 is given in Chapter 2.

6.1.2. As-Cast Aspects

The heterogeneities in structure of the semi-continuous (direct-chill) as-cast AA6063 alloy can be seen in Plates (6.1. a-c). Most of the magnesium is in solid solution, the remainder existing as a eutectic of Mg_2Si and $AlFeSi$ at the primary dendrite arms. The most prominent feature of the as-cast material is the residence of the coarse iron-bearing intermetallic in the grain boundaries. This phase is primarily present in the $\beta AlFeSi$ form which has a monoclinic structure.

The $\beta AlFeSi$ phase appears white in the SEM (BEI) micrograph (Plate (6.1.b)) and black in the TEM micrograph (Plate (6.1.c)). The appearance of black precipitates in the grain boundary shown in Plate (6.1. b) are in fact a mixture of coarse Mg_2Si (β) and voids where the $AlFeSi$ phase has been pulled out of the matrix during sample preparation and polishing. The latter being more frequent. To identify the Mg_2Si (without EDX), the SEM is switched from BEI to SEI mode during which the voids remain black.

The ' β ' iron phase in aluminium alloys presents a greater detriment to extrusion, on both final material properties and to the surface finish, than the more equiaxed ' α ' phase. One of the principal reasons for the advantage of $\alpha AlFeSi$ phase is that the transformation renders the brittle plate like $\beta AlFeSi$ phase into a less brittle, more

(a) As-Cast (Optical)

(b) As-Cast (SEM)

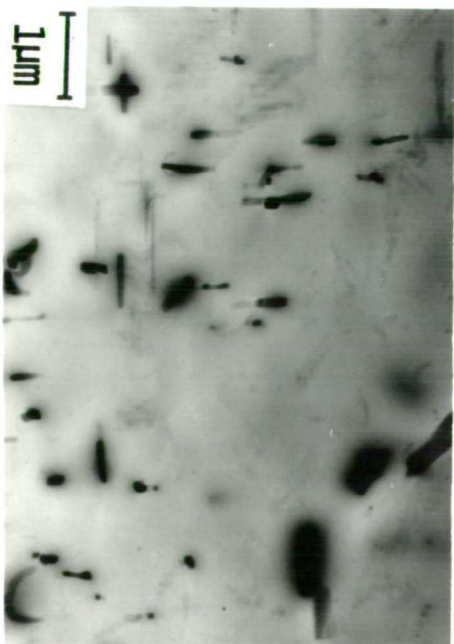
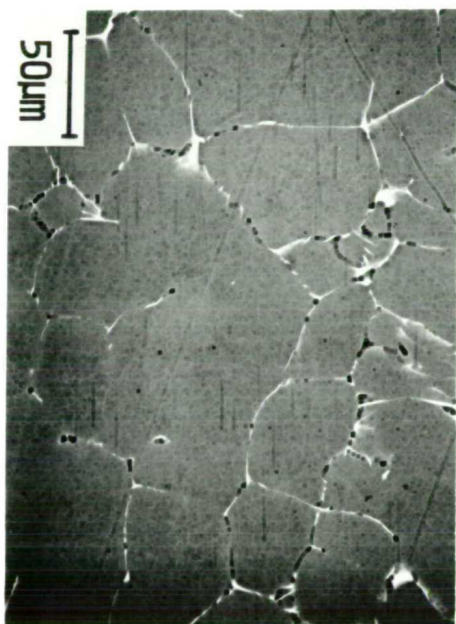
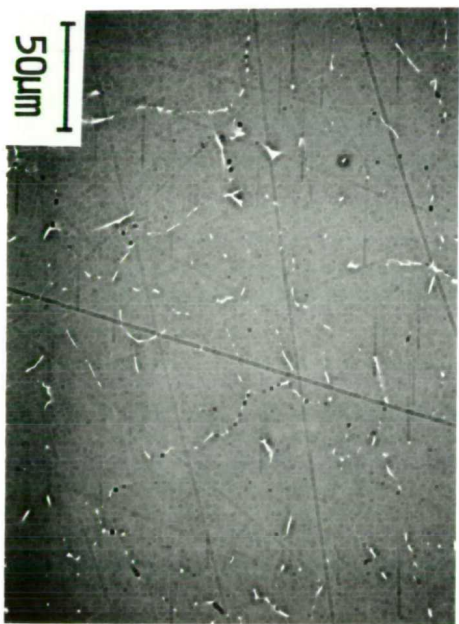
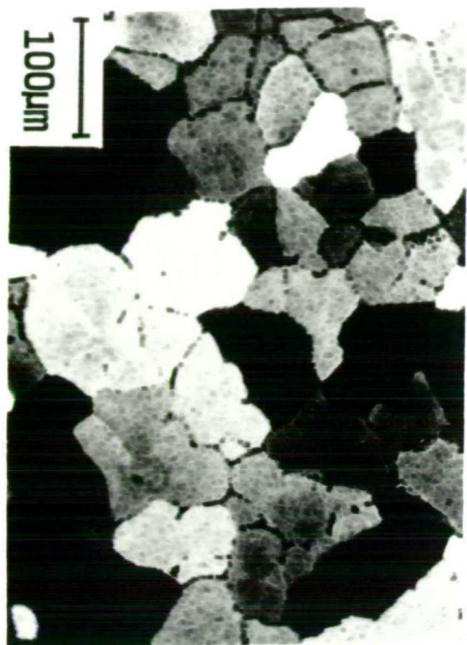
(c) As-Cast (TEM)

(d) Homogenised (Optical)

(e) Homogenised (SEM)

(f) Homogenised (TEM)

Plate 6.1. As-Cast and Homogenised AA6063 Structures.



rounded form^(42,127). Also, during the transformation the βAlFeSi phase loses some of its Si into solid solution which increases its availability to be precipitated at the age hardening Mg_2Si phase.

In view of these advantages, it is necessary to perform a homogenisation treatment, prior to extrusion, to obtain a superior extrusion product.

6.1.3. Homogenisation Treatment

Satisfactory homogenisation, for the purpose of this study, was not characterised by the attainment of complete equilibrium in the matrix structure. The major (industrial) criteria for homogenisation practices in AA6063 is high extrudability with good surface finish and adequate mechanical properties. It was this measure that was considered most pertinent in the present work. A (short industrial) homogenisation treatment was performed on the AA6063 alloy, of 6 hours at $575^\circ\text{C} \pm 5\%$ and cooled at 150°C/hr to 250°C then rapidly cooled to room temperature. The structure obtaining from this treatment can be seen in Plates (6.1. d-f). The heavy coring apparent in the as-cast structure has been considerably reduced (Plate (6.1. d)) and the intermetallic iron-phase has a more rounded appearance which is characteristic of $\alpha\text{AlFe(Mn)Si}$ (Plate (6.1. e)). Although the transformation from β to α occurs rapidly, especially in the presence of Mn⁽¹⁸⁾, the change of shape into equiaxed α is less rapid. The accelerative effects of Mn can be seen in Figure (6.1.) where additions of up to 0.1%wt. Mn drastically reduces the time for transformation to occur. The alloy used in the present investigation had small

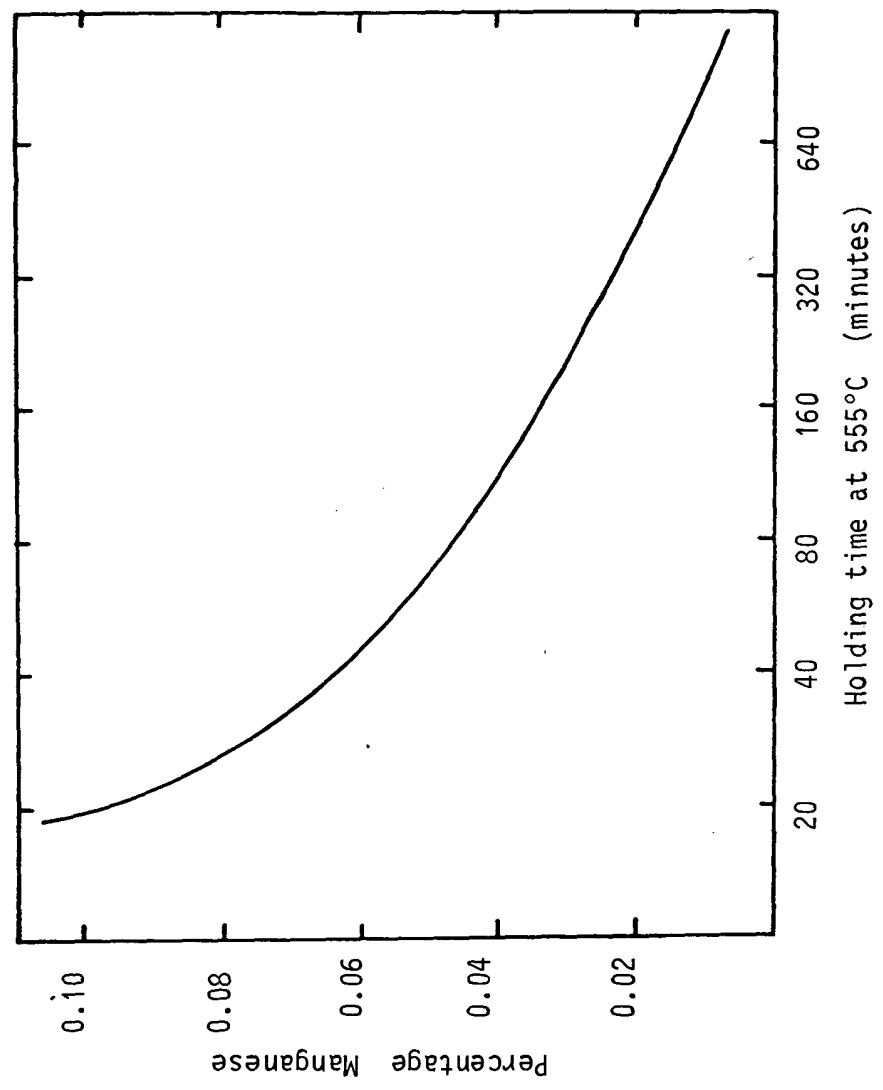


Figure 6.1. Effect of Mn Content on β to α Iron Phase Transformation

additions of Mn resulting in a required homogenisation time of 1 hour for transformation to occur. However, this time was increased to 6 hours for the desired globular form to result.

A fine distribution of the second phase Mg_2Si can be seen in Plate (6.1.f). The majority of the Mg_2Si is present in the meta-stable β' form, which are rod shaped, and lie in the cube direction; that is $\langle 100 \rangle$.

Also present in the homogenised structure is the equilibrium β (Mg_2Si) phase which are platelet shaped.

The homogenised structure shown in Plates (6.1.d-f) is quite satisfactory due to the iron phase transformation from β to α and the higher density of Mg_2Si in the β' form which more readily dissolves on passing through the deformation zone during extrusion, offering a combination of lower extrusion pressure and near complete dissolution of the second phase in the extrudate product.

It can also be seen in Plates (6.1. a & d) that there was no significant increase in the grain size during the heat treatment and remained at 90-100 μm .

6.2. Extrusion Product

The structural observations on the extrusion products were made at room temperature after rapid quenching from the high temperature deformation. These hot worked structures are unstable and tend to

recrystallise statically after deformation - if held at these temperatures - resulting in the loss of the sub-structure. This is more prevalent at the extrudate surface where the driving force for migration of high angle boundaries is fueled by the increased degree of deformation imposed on the material feeding the extrudate annulus.

The sub-structure, which has been 'quenched in', generally exhibits a dynamically recovered nature; one in which the original grains are elongated in the extrusion direction containing an interior of smaller sub-grains formed by a mechanism of generation and annihilation of dislocation pairs. The retention of sub-structure, although not generally commercial in AA6063, significantly contributes to its final mechanical properties. This feature is discussed in Section 6.3. In the following sections the effects of process conditions, mode of extrusion and the section position on the structural and sub-structural features will be discussed in detail.

6.2.1. The Effect of Process Parameters on Structure

Extrusion trials were performed in the 350°C to 550°C temperature range and at extrudate exit speeds from 30mm s⁻¹ to 4800mm s⁻¹.

Within this matrix of process conditions, three major types of structure were identified. The first consisted of a fibrous inner core, (Plate (6.2.a)) surrounded by an annulus of recrystallised grains (Plate (6.2.b)). This type of structure was typical of extrudates processed at the lower (350°C) temperature conditions, coupled with press water quenching. The fibrous structure shown in Plate (6.2.a) consists of the original homogenised grains, elongated

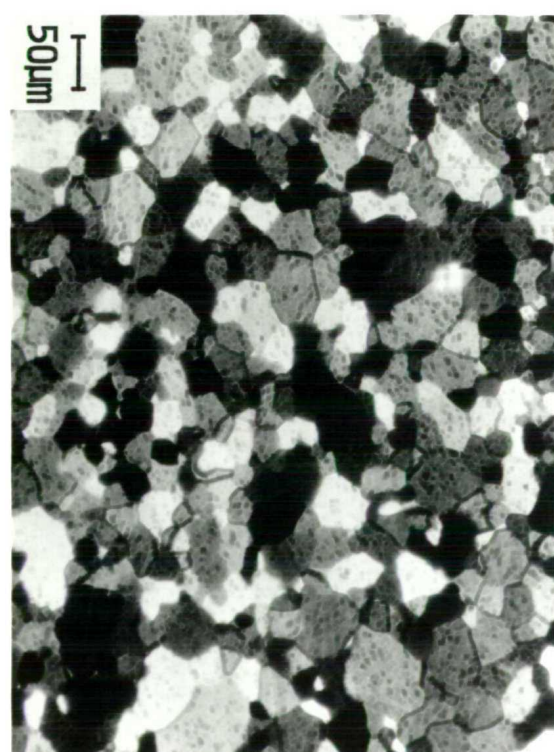
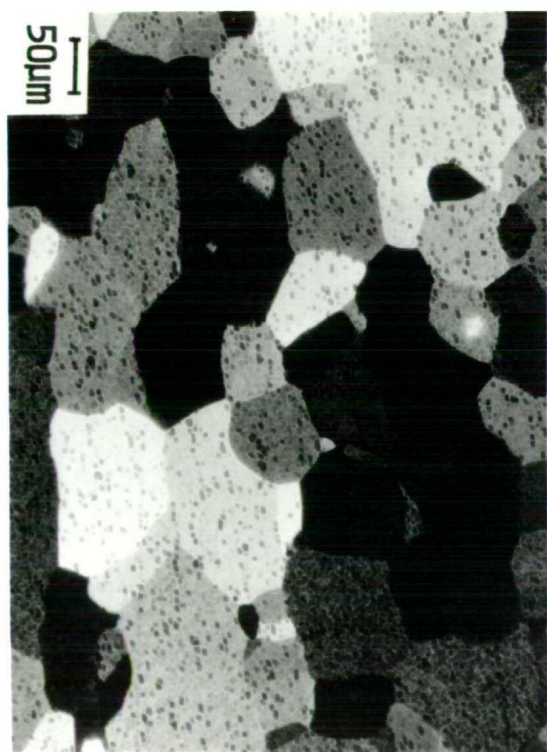
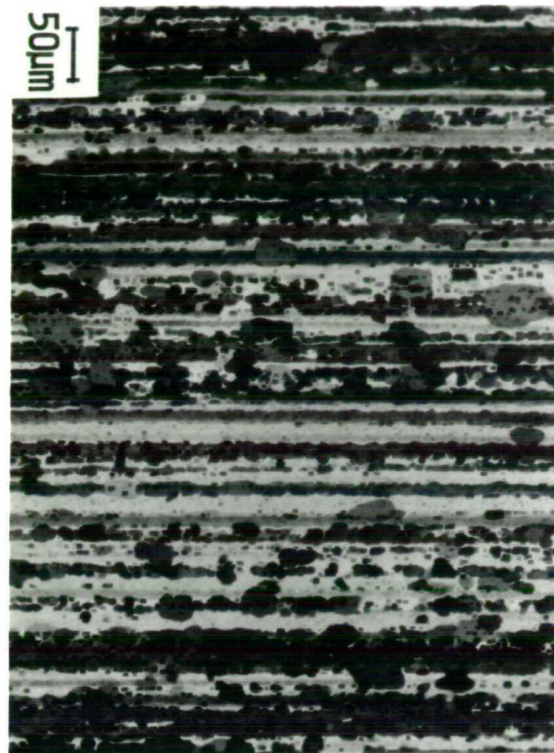
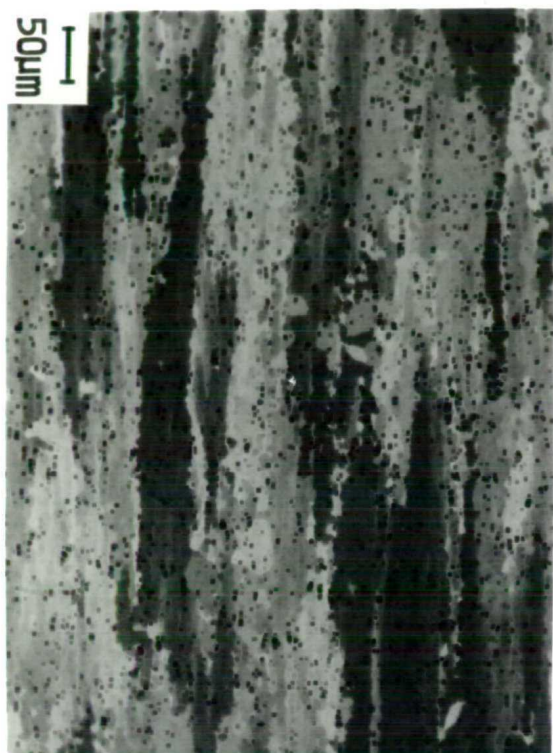
(a) Fibrous.

(b) Recrystallised Annulus.

(c) Elongated Recrystallised.

(d) Recrystallised.

Plate 6.2. Extrusion Structures.



in the extrusion direction. Though not apparent in the optical micrograph, transmission electron microscopy reveals the fibrous grains to contain an interior of smaller sub-grains. These low angled sub-grains were formed dynamically by a continuous rearrangement of dislocations at the sub-boundary walls which maintained their almost equiaxed form. Also, within the fibrous structure in Plate (6.2.a), occasional recrystallised grains can be seen which have formed statically.

However, as the thermal energy was relatively low and the time for grain growth limited, due to water quenching the recrystallised size was limited to 15 μ m to 25 μ m in size.

This type of 'duplex' structure in the core of high 'Z' extrusion is characteristic of dilute alloys such as AA6063 and commercial purity aluminium, when water quenched after extrusion.

The metal feeding the annulus originates from the heavily sheared regions flowing from the front portion of the deformation zone, adjacent to the dead metal zone. The material from this region has a high dislocation density due to the increased deformation, producing a greater number of nucleation sites and internal energy which is sufficient to promote recrystallation. The high density of nucleation points will, relatively, impede grain growth resulting in a typical recrystallised grain size of approximately 18 μ m in the annulus at high 'Z' conditions (Plate (6.2.b)).

The second type of extrusion structure had an elongated appearance but with a much lower aspect ratio than that of the fibrous structure.

This type of structure is termed 'elongated-recrystallised' and occurred in the mid-temperature range; 400°C (Plate (6.2c)). The increased thermal energy has permitted static recrystallisation to a greater extent. The alignment in the extrusion direction of the coarse iron phase has impeded grain growth in the transverse direction, encouraging the motion of high angle boundaries in the extrusion direction, resulting in the elongated appearance. The formation mechanism of this type of structure is via the Zener drag phenomenon which has been recently reviewed by Nes et al⁽¹²⁸⁾. The annulus was again made up of small equiaxed recrystallised grains, which were slightly larger than the previous condition as a consequence of greater recovery during deformation and its associated decrease in nucleation sites, coupled with greater thermal energy for grain growth.

The third type of structure was one of complete recrystallisation (Plate (6.2.d)). This structure was generally observed at high temperatures (>500°C) with water quenching and at all hot working temperatures, and higher ram speeds, with air quenching. This structure resulted from the mechanism of static recrystallisation and is one which is (industrially) most common in extrusion of AA6063.

The practice of water quenching the extrudates, immediately as they emerge from the die, was adopted so that the hot worked sub-structure might be retained to assist the interpretation of the mechanisms operative during deformation. The use of optical micrographs to characterise the types of extrusion structures obtained can only give a coarse representation of micro-structure. Although this is of actual importance the true nature of structure may only be observed at

greater magnifications available in the TEM.

6.2.1.1. Sub-Structural Detail

Specimens for electron microscopy were taken from equivalent positions (with respect to the ram stroke) along the length of each extrudate and prepared using the methods described in Chapter 4. Samples were only taken from that part of the extrudate which coincided with steady state extrusion.

The steady state structures of the water quenched extrudates all exhibited a portion of 'quenched in' sub-grain structure, the proportion of retention increasing with decreasing temperature conditions. This results from the reduced thermal energy available for static recrystallisation to occur. The low contrast between grains and the small angle of rotation of the diffracted electron beam on traversing the boundary, indicated that a low angle misorientation existed between the grains, which is consistent with the operation of dynamic recovery and repolygonisation during extrusion. Further confirmation of a low angle boundary was obtained by tilting the specimen foil in the TEM to observe the dislocation network present at such interfaces. The remainder of the extrudate structure was statically recrystallised, displaying a relatively dislocation-free interior, resulting from the simultaneous annihilation of a large number of dislocations from the migration of a high angled boundary. There was no evidence of dynamic recrystallation in the extrusion microstructure of AA6063. This was further strengthened when considering the stress/strain curves produced in hot torsion. These

curves showed no dramatic drop in flow stress after peak stress, or oscillations in the 'steady state' stress, which are both characteristics of dynamic recrystallation.

In Plates (6.3.a-c) examples of sub-structure obtained during low temperature (i.e. high Z) extrusion can be seen. The sub-grains are elongated in the extrusion direction and have a relatively high internal dislocation density. This indicates that there was insufficient thermal energy at low temperature (350°C), high Z, to allow dynamic recovery to complete. Recovery is reduced further with increasing ram speed.

There is generally a large variation in sub-grain size, and some evidence of sub-grain banding due to containment in the extrusion direction (Plate (6.3.b)). The absence of second phase Mg_2Si precipitation is also apparent in Plates (6.3 a-c). This is characteristic of the AA6063 extrusion product as it has a low quench sensitivity, allowing the retention of most of the magnesium and silicon in solution. (This retention of solute after extrusion allows tempering and ageing without the need for a separate solution treatment. Although not apparent in Plates (6.3. b & c), elsewhere in the sample foil, there was an alignment in the extrusion direction of the ever present intermetallic $AlFeSi$ and some coarse Mg_2Si ; these coarse precipitates being elongated in the extrusion direction. The coarse precipitates can be seen to occasionally pin the sub-grain boundaries, Plate (6.3. a). This is not a general feature as the dislocations in the sub-grain boundary more often possess adequate energy from the applied stress to surmount the relatively low percentage of incoherent precipitates presenting obstacles in this

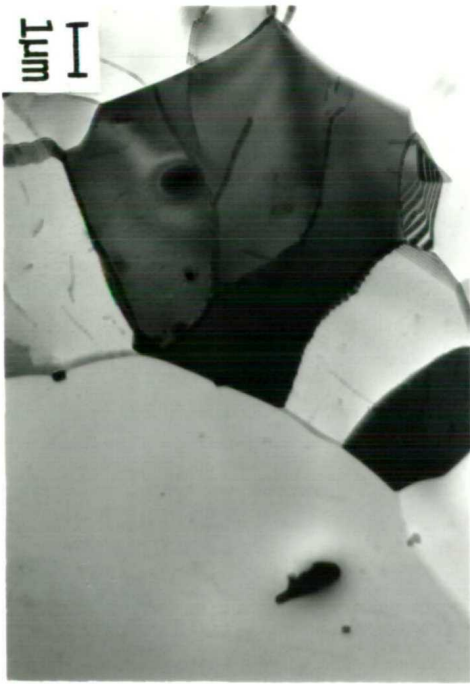
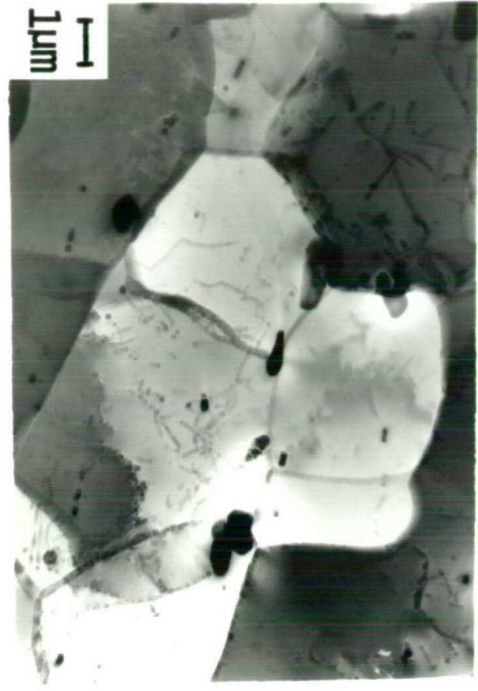
(a) High Z

(b) High Z

(c) High Z

(d) Medium Z

Plate 6.3. Extrusion Structures viewed in TEM.



alloy. Consequently, these precipitates do not appear to govern the sub-grain size.

When the extrusion temperature is raised ($>450^{\circ}\text{C}$) there is an increase in both sub-grain size and perfection, as shown by the narrower sub-boundary walls and reduced internal dislocation density. This is evident in micrographs shown in Plates (6.3. d and 6.4. a). The sub-grains shown in Plate (6.3. d) are adjacent to a high angle boundary of a statically recrystallised grain. The increased size and perfection result from the increased thermal energy aiding both cross-slip and climb of the dislocations which increases the mobility of the sub-boundary dislocations allowing the cell walls to migrate and form more equiaxed sub-grains. The driving force for the cell wall migration is associated with the minimisation of the total grain boundary area, which reduces the free energy of the structure.

As the extrusion temperature is raised further ($>500^{\circ}\text{C}$) there is sufficient thermal energy to result in an almost totally recrystallised structure. The large recrystallised grains exhibit a dislocation-free interior and sharp boundary walls and so have formed statically from the migration of a high angled boundary. In Plate (6.4. b) a triple point of recrystallised grains can be seen. Occasionally, the migration of the boundary is arrested as it is pinned by the larger AlFeSi precipitation. In Plate (6.4. c) an AlFeSi precipitate which has been elongated in the extrusion direction can be associated with grain boundary pinning. Static recrystallisation in the AA6063 alloy is not confined to the high temperature extrusions and will occur even at the low temperature of 350°C . This is especially prevalent when air quenched, as opposed to water quenched after deformation. In Plate

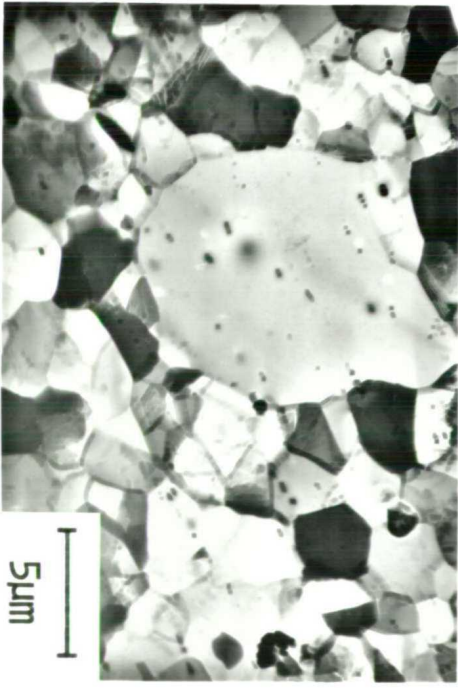
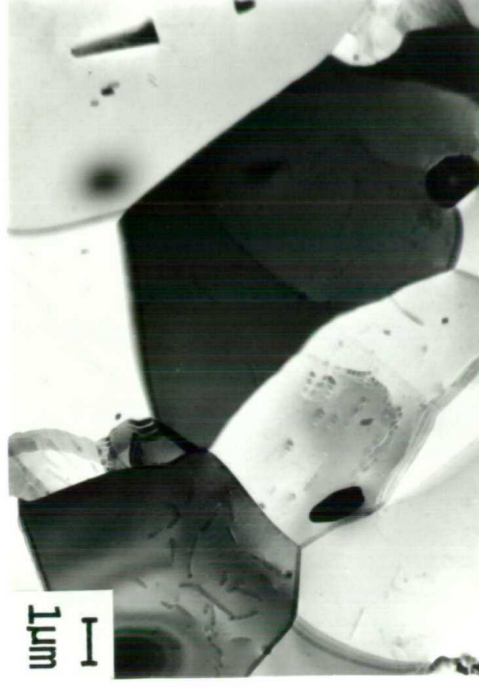
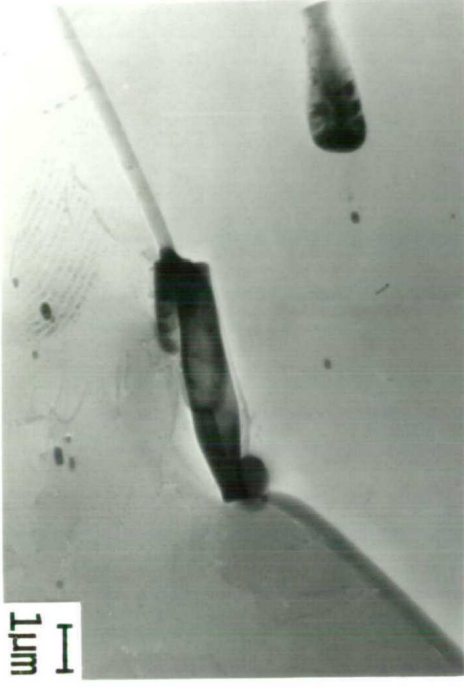
(b) High Z

(a) Medium Z

(d) Static Recrystallisation

(c) Boundary Pinning

Plate 6.4. Extrusion Structures viewed in TEM.



(6.4.d) a statically recrystallising grain can be seen migrating through the sub-grains. The low temperature extrusion conditions produced the fine sub-grain structure but the water quench was not rapid enough to prevent recrystallisation beginning.

6.2.1.2. The Effect of Process Conditions on the Steady State Sub-Structure

In the previous section it was shown that the sub-grain size was dependent upon the processing condition; increasing in size with temperature and, to a lesser extent, with decrease in strain-rate. It would therefore appear reasonable that a relationship between the sub-grain diameter (d) and the temperature compensated strain-rate (Z) should exist. It has been shown, for a number of aluminium alloys, that a relationship of the form:-

$$d^{-m} = a + b \cdot \ln(Z) \quad (6.1.)$$

gives good correlation. In this equation, a , b and m are constants with ' m ' taking the value of 1.

To explore the applicability of equation (6.1.) to the hot deformation of AA6063, the 'mean linear intercept' method was used to measure the sub-grain size, produced at varying process conditions, for both extrusion and hot torsion tests. Due to the variance of sub-grain size at each temperature/strain-rate pair, the mean of at least 100

readings were taken. The value of the temperature compensated strain-rate was corrected for the temperature rise produced during deformation using the integral profile method described by Cooper⁽¹⁴⁾. The results from this study are shown in Figures (6.2. a-b) and the best fit equations obtained from these results were:-

$$d^{-1} = 0.025 \ln(Z) - 0.455 \quad \text{cc-0.936 (for direct)}$$

$$d^{-1} = 0.019 \ln(Z) - 0.391 \quad \text{cc 0.944 (for indirect)}$$

$$d^{-1} = 0.016 \ln(Z) - 0.230 \quad \text{cc 0.971 (for torsion)}$$

These equations are consistent with those obtained by other workers (7-9,11-14), although this alloy's sub-grain size is not as sensitive to varying process conditions, as indicated by the comparatively low values for 'a'. This is demonstrated graphically in Figure (6.2. c). The reason for this is the high rate of recovery operative during deformation of AA6063, resulting from its dilute composition. In the higher strength alloys AA2014⁽⁹⁾ and AA5456⁽⁷⁾ the resultant sub-grain size is smaller than that of AA6063 (Figure (6.2.c)) due to the higher percentage of second phase precipitates hindering sub-boundary motion. In addition, the greater solute percentage in these alloys tends to retard dynamic recovery, as shown by McQueen⁽¹³⁰⁾ for magnesium alloys. Hence the difference in the sub-grain's size produced from AA5083⁽¹⁴⁾ and AA5456⁽⁷⁾ is partly a result of the higher solute content in AA5456. It can also be seen in Figure (6.2.c) that the sub-grain size produced is comparable with that of AA1100 aluminium from the work done in the hot rolling of commercial purity aluminium by Zaidi⁽¹³¹⁾.

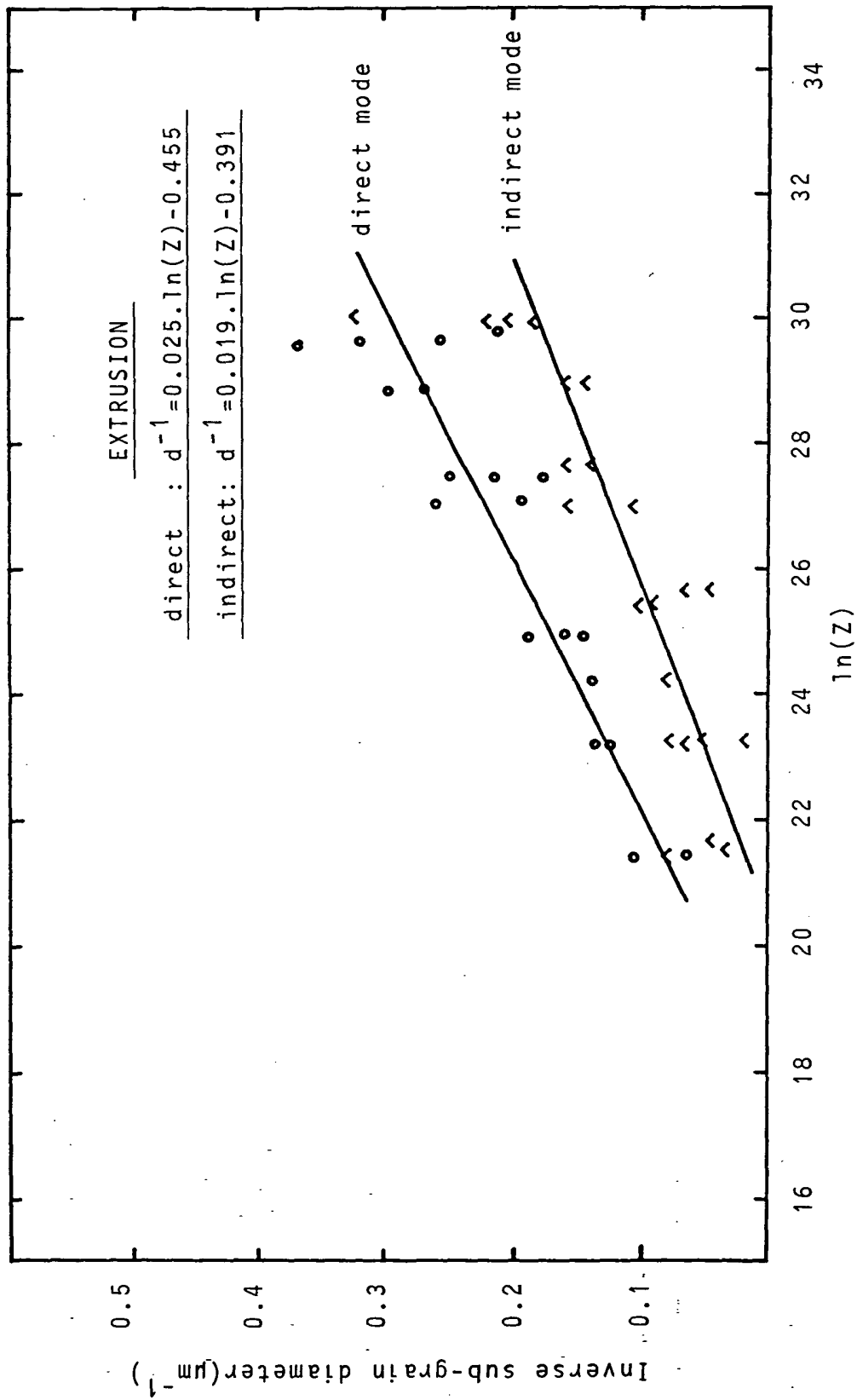


Figure 6.2 (a) Extrusion

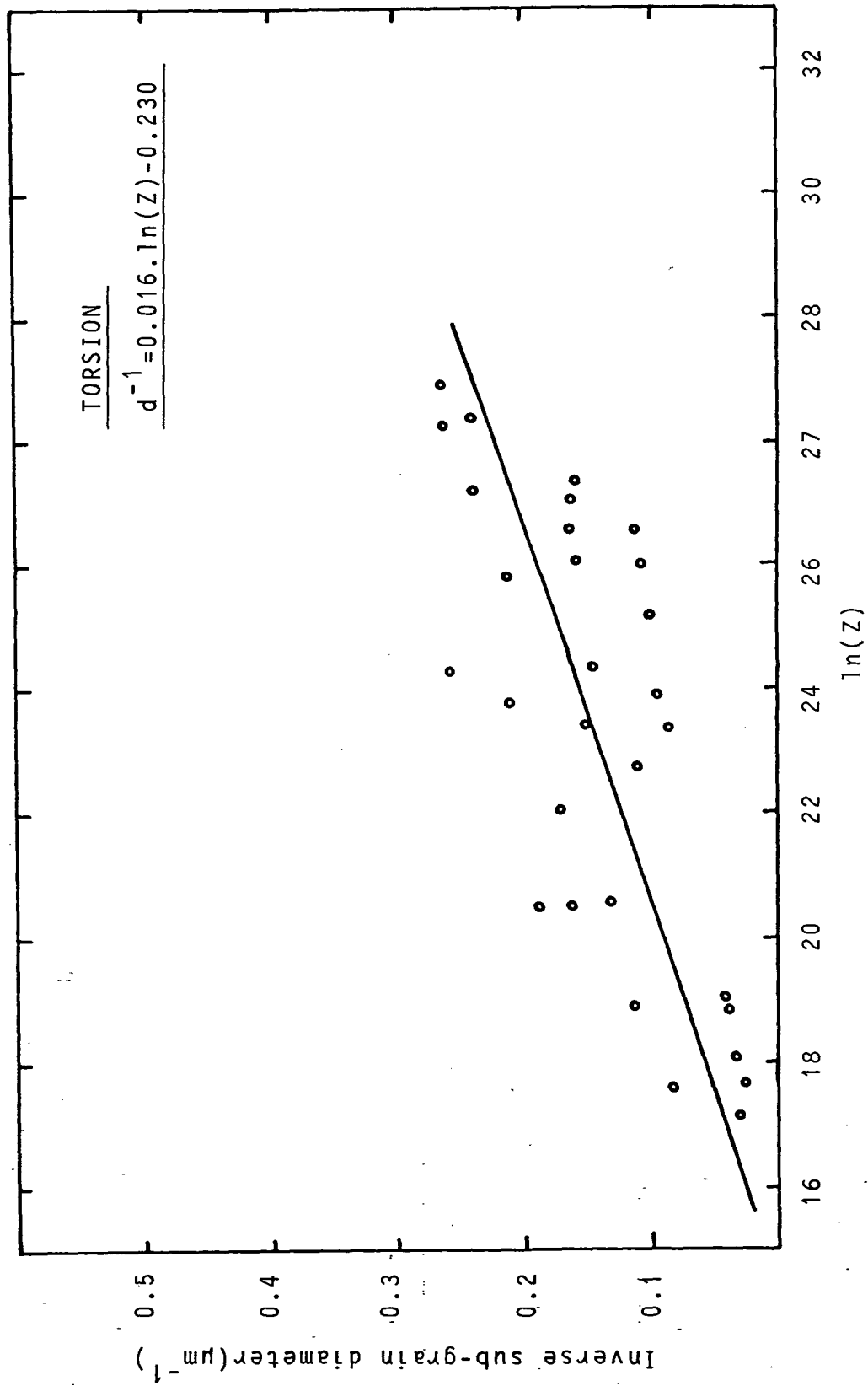


Figure 6.2. (b) Torsion

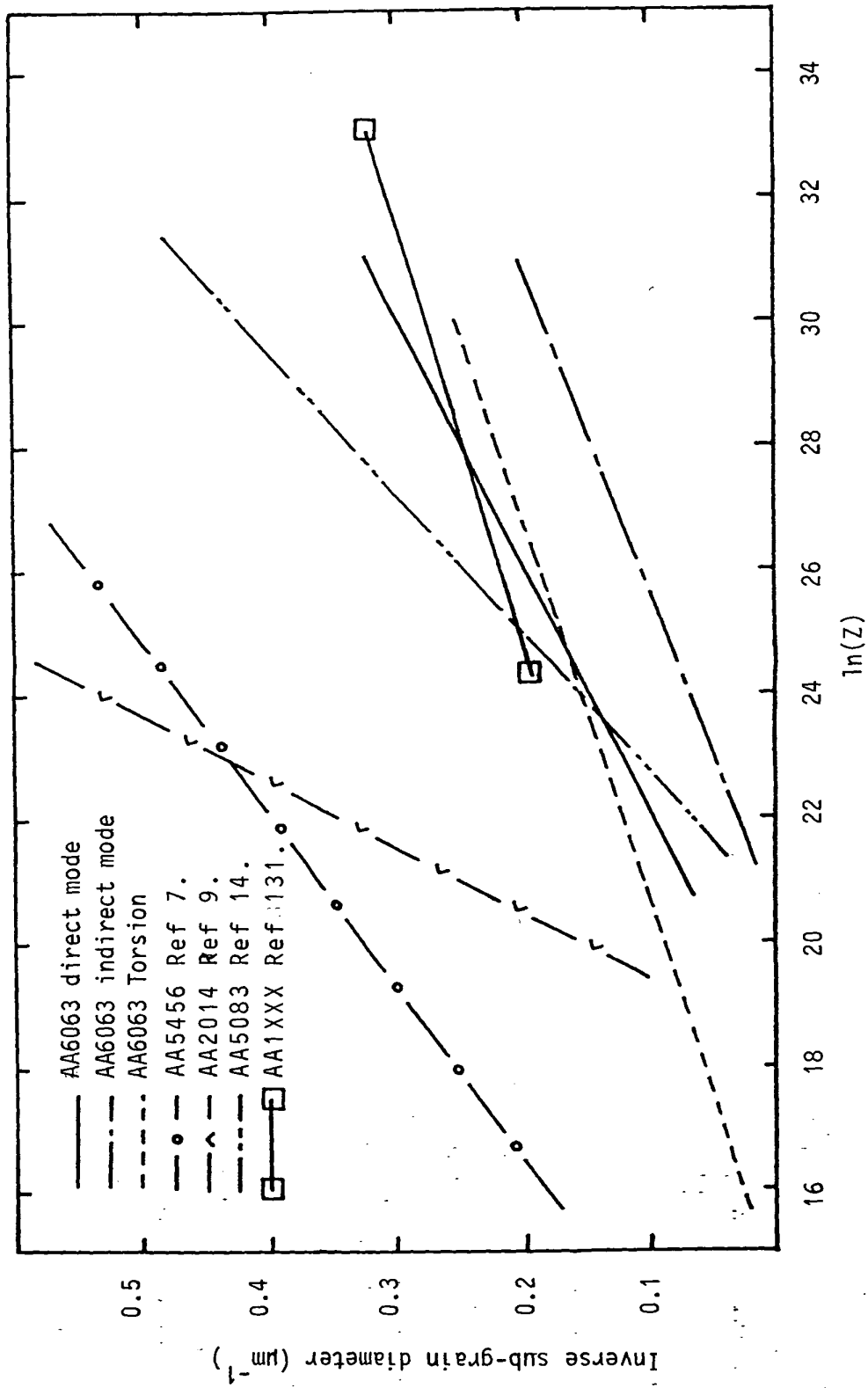


Figure 6.2. (c) Comparison

Another major feature reflected in Figure (6.2.a) is the non-coincidence of the 'best fit lines' through the data from direct and indirect extrusion, and hot torsion. The departure between extrusion and torsion data can be explained by considering the inherent inaccuracies in the analytical conversion of process data into effective strain-rate and temperature data. This should, however, not be so between direct and indirect extrusion data, and at identical temperature, strain-rate pairs should be coincident. A possible reason for this departure is sub-grain coalescence in indirect extrusion resulting from the quench delay imposed by the tooling arrangement (Figure 4.1.).

In order to explore this further, two billets were extruded at $\ln Z \approx 30$. The first trial was conducted at 350°C and a strain-rate of 14.4s^{-1} and the second at 300°C and a strain-rate of 1.3s^{-1} . Results indicated that the former trial produced a smaller average sub-grain size of $3.1\mu\text{m}$ compared with $5\mu\text{m}$ at 300°C . This is consistent with sub-grain coalescence but is by no means conclusive evidence for the departure of the result in Figure (6.2.). However, this trial indicates that further research is necessary in this area. The use of a mean strain-rate may be in error for the indirect process. It was shown by Sheppard and Paterson⁽¹²⁹⁾ that a rapid increase in the strain-rate close to the die entry affects the structure significantly and concluded the use of a mean strain rate inadmissible.

However, bearing in mind these reasons for departure of coincident equations relating sub-grain size and temperature compensated strain rate, it is evident that there is a strong similarity between each

deformation process and confirms the validity of using the hot torsion test as a model for extrusion, so long as caution is exercised in the interpretation of the results.

6.2.2. The Effect of Direct and Indirect Extrusion on Structure

In general terms, the extrusion process involves shaping a material by forcing it through a suitably shaped orifice, and so geometrically, both direct and indirect extrusion can be considered identical. However, on closer examination, the material flow during extrusion and the development of structure are quite different. The disparity between the material flow, in each extrusion mode, results from the differing friction condition pertaining. Plate (6.5.) shows the contrast between the two types of flow using super purity (SP) aluminium.

In the direct process, a relative movement between the billet and container exists encouraging inhomogenous flow and as a consequence, localised heating occurs, and a temperature gradient across the extrudate section results. The extrudate surface also suffers a higher degree of deformation and can result in a (sometimes) deleterious recrystallised annulus. The region of intense shear feeding the extrudate surface can be distinguished clearly in Plate (6.5.a) extending from the die face, adjacent to the die mouth, to the container/billet interface located towards the rear of the billet. This deformation zone is almost hemispherical in shape and decorates a portion of stagnant metal at the front of the billet called the dead

(a) Partial extrusion (direct mode)

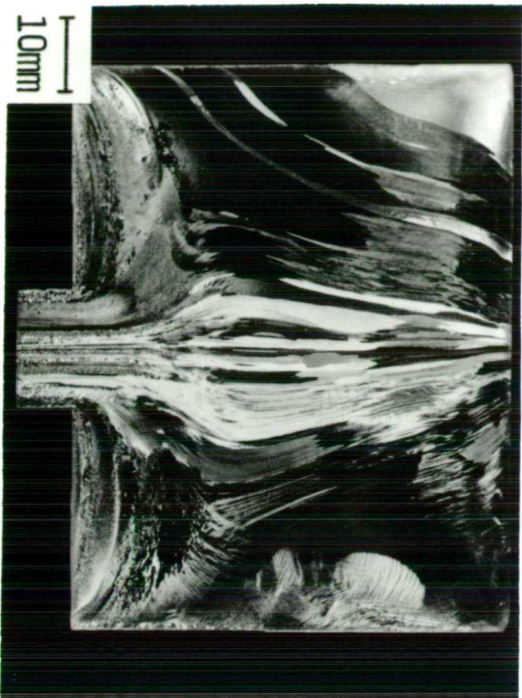
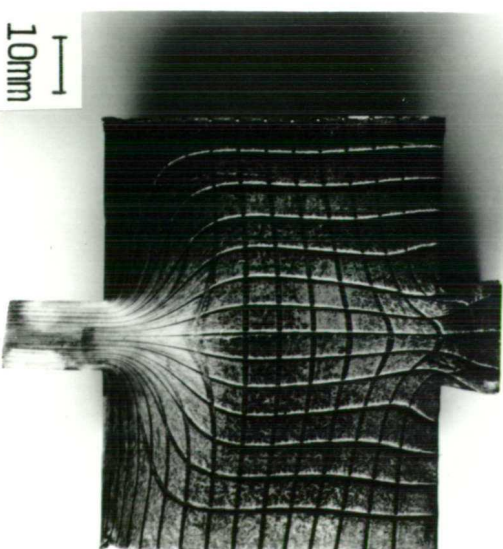
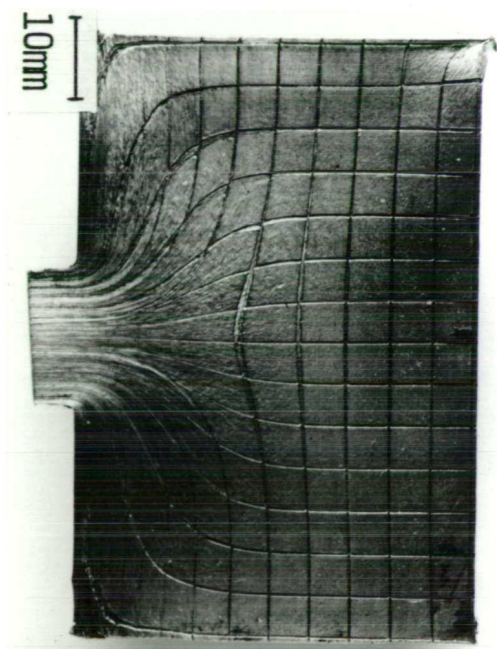
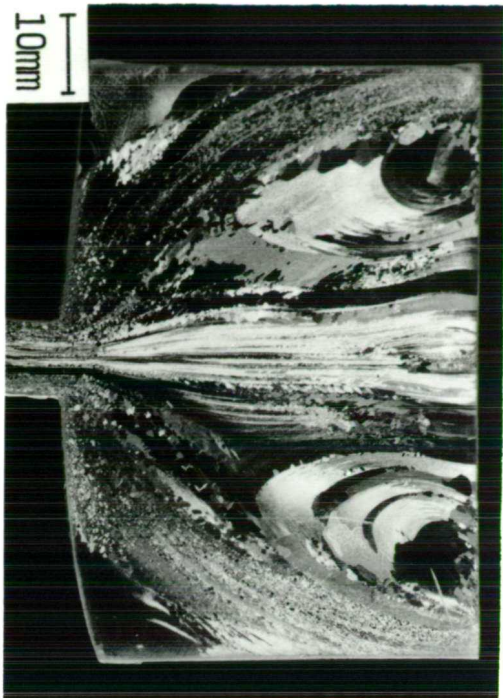
(c) Low temperature (direct)

(e) Medium temperature
(indirect)

(b) Partial extrusion (indirect mode)

(d) High temperature (direct)

Plate 6.5. Extrusion Flow Patterns.



metal zone (DMZ). Conversely, in the indirect process, a stationary billet/container interface eliminates the frictional contribution and results in a more homogenous process. It is noticeable in Plate (6.5.b) that the shearing is not as intense as that observed during direct extrusion. The deformation zones feeding the extrudate are apparent adjacent to the die face and in the centre region.

During indirect extrusion the billet surface and its accompanying defects and heterogeneities, flow into, and form part of the extrudate surface. Hence, it is recommended practice that extrusion billets be scalped - especially aerospace alloys - prior to processing. However, the surface generation is via the same shear mechanisms as those operative during direct extrusion but are of a lesser intensity. Thus a more homogenous product is obtained, both along and across the extrudate, during indirect extrusion. The heavy shearing evident at the rear of the billet is caused by metal flow into the dummy block and is not a consequence of the indirect process.

In Plate (6.6.) comparisons are made between structures produced via the two extrusion modes (direct and indirect) at low, medium and high Z (550°C, 450°C and 350°C respectively) process conditions, using a constant ram speed of 14mm s^{-1} and an extrusion ratio of 20:1. Specimens were then taken from the centre of the extrudates. The location of the specimens along the extrudate corresponded to quasi-static deformation at a position 0.5 of the ram stroke. The most outstanding contrast between the structures obtaining from the two extrusion modes was observed at an extrusion temperature of 350°C (high Z). (Shown in Plate (6.6.a,b)).

The structure obtained from the direct process had a fibrous appearance speckled with statically recrystallised grains (Plate (6.6.a)) whereas the indirect case displayed a more highly recrystallised structure with a slight aspect towards the extrusion direction (Plate (6.6.b)). The difference in structure is most probably because the total stress and strain-rate in direct extrusion tend to produce a greater number of nucleation sites which will, relatively, impede grain growth. It was also found that the intermetallic iron phase suffers less distortion and elongation, in the extrusion direction, in the indirect mode. Thus containment in the extrusion direction of the recrystallised grains is less likely⁽¹²⁸⁾.

Another major factor governing the structural differences was the quench delay, imposed by the tooling arrangement, for indirect extrusion. This delay (one second in the case of these trials) was sufficient to enhance sub-grain coalescence and static recrystallisation. An increase in extrusion temperature (to 450°C) results in greater static recrystallisation in both modes of extrusion. The directly extruded structure continues to exhibit a directionality in its recrystallised grains (Plate (6.6. c)). However, this is far less apparent. The grain size is smaller than the equivalent indirect structure as a result of a combination of increased nucleation sites and reduced time - due to quench arrangements - for grain growth. The indirect structure exhibits no directionality other than in the alignment of precipitates, (Plate (6.6.d)). At the highest extrusion temperature (550°C), total static recrystallisation has occurred. The recrystallised grain size produced via the direct mode of extrusion, at this temperature, was 60µm.

(a) High Z (direct)

(b) High Z (indirect)

(c) Medium Z (direct)

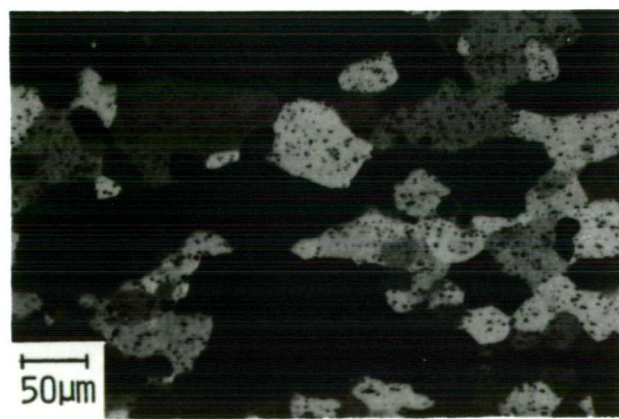
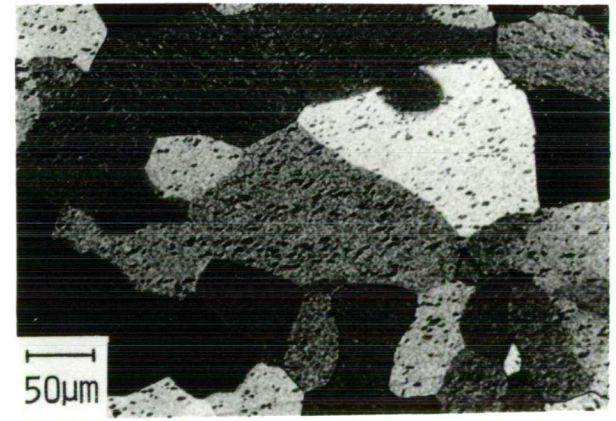
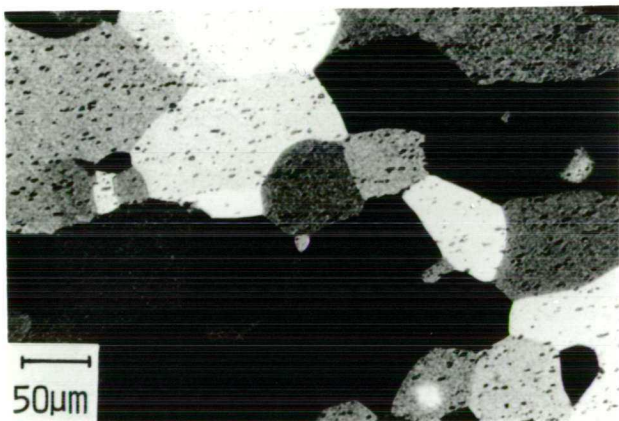
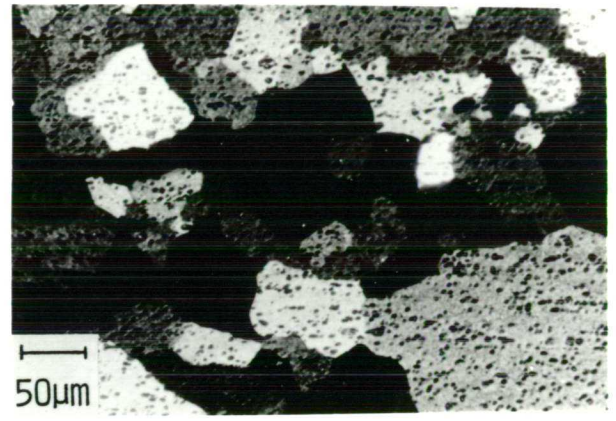
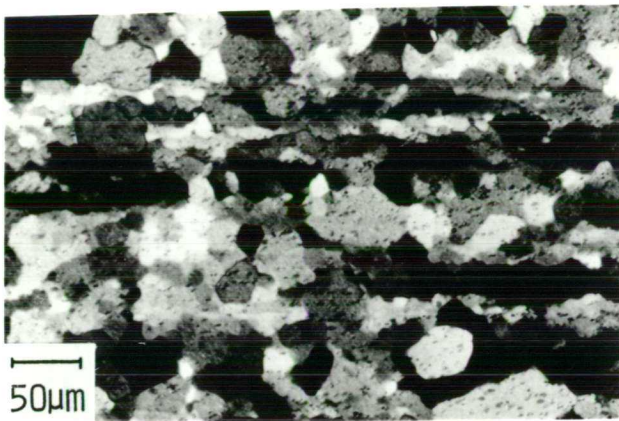
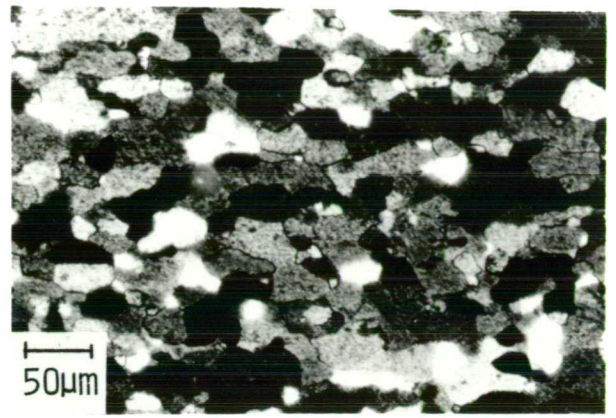
(d) Medium Z (indirect)

(e) Low Z (direct)

(f) Low Z (indirect)

(g) Medium Z (indirect)

Plate 6.6. Extrusion Structures from direct and indirect modes (Optical)



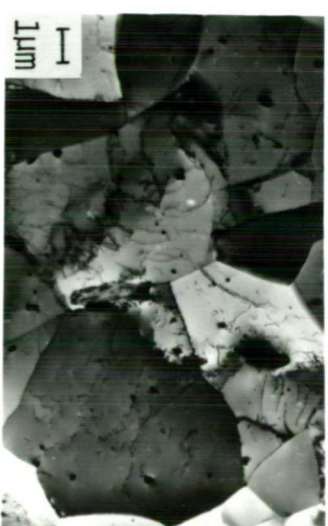
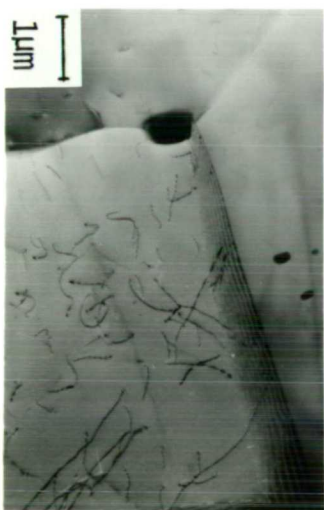
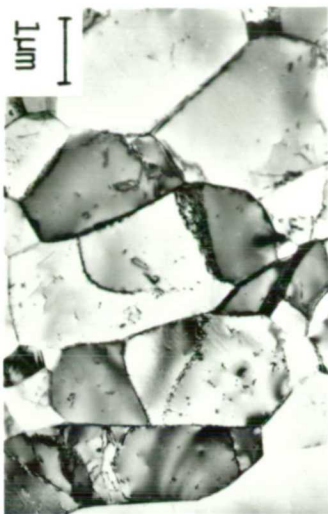
compared with 75 μ m in the indirect mode (Plate (6.6. e & f)).

This further strengthens the proposal that a reduced number of nucleation sites permits a larger recrystallised grain size. In order to ascertain the affect the quench delay had on the indirect structure, a further extrusion trial was performed. The trial conditions were such that the temperature compensated strain-rate was the same as that for the 450°C conditions (Plate (6.6.c,d)), but at a faster extrudate exit speed. This was achieved by increasing the initial billet temperature to 475°C which enabled the strain-rate to be increased by 120%, thus reducing the quench delay. The structure produced is shown in Plate (6.6. g). The grain growth has been considerably reduced and holds stronger similarities with the directly extruded structure (Plate (6.6. c)). However, as the quench delay for this trial was only reduced and not made equal to that in the direct mode, there is a recurrence of the coarser structure in the indirect extrudate, though to a lesser extent than that shown in Plate (6.6. d).

Transmission electron micrographs of the structures shown in Plate (6.6.) can be seen in Plate (6.7.). At high Z conditions the sub-grains are elongated with a relatively high dislocation density. The sub-grain size was slightly larger than that in the direct extrudate coupled with greater internal order of the dislocations. Other areas of the high Z specimens were found to contain statically recrystallised grains which is consistent with the results from optical microscopy. There does not appear to be any significant difference in morphology between the structures formed by direct and indirect extrusion: the only difference is the larger sub-grain size

- | | | |
|-----------------------|-------------------------|------------------------|
| (a) High Z (direct) | (b) High Z (indirect) | (c) High Z (torsion) |
| (d) Medium Z (direct) | (e) Medium Z (indirect) | (f) Medium Z (torsion) |
| (g) Low Z (direct) | (h) Low Z (indirect) | (i) Low Z (torsion) |

Plate 6.7. Comparison of structures produced by torsion and both modes of extrusion.



observed in the indirect extrudates. As extrusion temperature is increased to 450°C both modes of extrusion produce more equiaxed sub-grains with a lower internal dislocation density (Plate (6.7.d & e)). Those dislocations present have formed into well ordered cell walls (Plate 6.7. d). The percentage of retained sub-grains was less in the indirect extrudate which again is consistent with optical observations. The totally recrystallised structures produced at 550°C are shown in Plate (6.7.g & h). Although not a general feature, both modes of extrusion displayed high angled boundary pinning by coarse intermetallic iron phase.

6.2.3. Torsion Structure as a Model for Extrusion

In Chapter 5 it was discussed how the use of the hot torsion test could be used to model extrusion both structurally and constitutively. To compare the structures produced during torsion with those at equivalent 'Z' in extrusion, some thought must be given to the position of sample extraction from the torsion specimen. In a solid torsion specimen the strain, and stress rates vary from zero at the centre to a maximum at the surface, and so the radius at which the specimen is removed is important. Numerous workers^(5-9,11-15) who have used the hot torsion test to model steady state extrusion structure, have claimed to have extracted samples from the torsion specimen's surface, although at best, samples at 0.95 of the radius can only be taken. This is further reduced when surface roughening occurs with shear strain and the accompanied inhomogenous distortions at the surface. In the present study the use of the 'effective radius' method to convert torque/twist data into effective stress/strain data, it was considered appropriate

to extract specimens from this radial position, with corresponding calculations for temperature (corrected for temperature rise) and strain-rate.

The torsion specimens were machined from homogenised billets and therefore had the same starting structure as that shown in Figure (6.1. d-f) for extrusion. The torsion specimens were heated at $100^{\circ}\text{C min}^{-1}$ by induction and maintained at the test temperature by a Eurotherm controller. On completion of the test, the specimen was automatically water quenched so as to retain the hot worked structure. A comparison of the structure obtained from torsional deformation and both modes of extrusion is made in Plate (6.7.). As expected (and demonstrated in Section 6.2.1.2.) the structures have similar characteristics at equivalent temperature compensated strain-rates.

Although the structure is strain independent after attaining steady state flow stress, the strains induced on the torsion specimens were similar to those imposed in extrusion. This was to produce a comparable elongation and alignment of the intermetallic iron phase as a precaution to this phase affecting boundary and sub-boundary motion. However, on further examination of the sub-structure using electron microscopy, it was evident that this influence was negligible.

A major departure in the comparison of structures produced in torsion and extrusion was the absence of a recrystallised annulus equivalent to the duplex structure observed in direct extrusion at high Z processing. This is indicative of the much higher strains imposed on material feeding the extrudate annulus (and associated temperature rise) compared with the smaller strains and temperature rises.

predicted during torsional deformation.

Torsional deformation exhibits a higher dislocation density to either of the extrusion modes, but has a similar magnitude in sub-grain size. This is because the torsion specimens are quenched directly after deformation, and so static rearrangement of dislocation and recrystallisation is less likely to occur, unlike extrusion where there is a short delay in reaching the quench. In addition, the quench rate is faster in the hot torsion test than in extrusion. A feature, evident at medium and low Z , shown in Plate (6.7.f & i) is the presence of Mg_2Si precipitates. The probable cause for this is twofold. First, the heating rate prior to deformation is slower in the hot torsion test. This increases the time spent in the $275^\circ C$ - $425^\circ C$ temperature region where Mg_2Si rejection is most rapid⁽³⁶⁾. In addition, the heat generation during deformation may not be sufficient to allow the β' present in the starting structure to go into solution.

The resemblance between the sub-structural features shown in Plate (6.7.a-i) provide further confirmation that the use of hot torsion testing to model extrusion is valid. However, it is also clear that caution must be exercised when drawing parallels, carefully considering the limitations.

6.2.4. Structural Variations in the Deformation Zone and Effect of Process Parameters

To investigate the mechanisms involved in the development of sub-structure, billets were partially extruded. The point at which the

extrusion trial was arrested coincided with quasi-static deformation indicated by the pressure/displacement curve reaching steady state; that is, $dP/dx=0$. The 'partials' were then extracted from the container and rapidly quenched. The time from cessation of the trial to water quenching was approximately 120 seconds and so some static recrystallisation was inevitable. This was especially so at high temperature; low Z. This is quite evident in Plate (6.8.c).

Before considering the structural variation in the deformation zone and the associated effect of the process conditions, it is useful to explore the deviations in the macroscopic flow patterns of extrusion at high and low values of temperature-compensated strain-rate.

This was achieved by sectioning billets along a meridian plane and machining a square grid on one half of the billet. After each cut surface was treated with graphite lubricant (to prevent welding during extrusion), the halves were rematched and spot welded prior to extrusion. The prepared billets were then partially extruded, removed from the container and quenched. These were subsequently prised open to display the distorted grid pattern. In Plate (6.5.) flow patterns produced at high and low temperature conditions can be seen and it is evident that there is a significant change in deformation zone shape.

At the low temperature condition, (Plate (6.5.c)) the deformation is strictly limited to a small volume close to the die mouth, leaving approximately 50% of the rear-remaining billet relatively undeformed. This is indicated by the retention of the square grid. The formation of a DMZ is evident at the front of the billet reaching back from the die throat to the container wall at an angle of approximately 25° . In

comparison, the high temperature condition (Plate (6.5.d)) shows a considerable difference. In this event deformation is present over the total area exposed. There is a considerable amount of interface shearing between the container wall and extrusion billet stretching from the rear of the billet to the DMZ. The angle of this boundary has increased considerably giving a DMZ angle of approximately 50° .

The difference in the extrusion flow patterns is most suitably explained by considering the material's plastic flow behaviour. In the high temperature case, the material is much less able to withstand the strain gradient imposed by the deformation, but at lower temperatures the material has a greater resistance to plastic flow permitting the strain gradients to be contained in elastic strain. In view of the contrasting flow patterns, it would be expected that the structural development possesses a dependence on process parameters.

To study the structural development along flow lines traversing the deformation zone and at locations outside the intense shearing regions, a series of 'partial extrusion' trials were performed. During the extrusion cycle the process was stopped when the extrusion had reached quasi-static deformation, similar to the gridded billet extrusion trials. The billets were subsequently removed from the container and rapidly quenched. The partial was then sectioned along a meridian plane using a cut-off wheel at a low feed rate and an application of a high flow rate of coolant. One half was then ground and polished to $14\mu\text{m}$ diamond paste and subsequently etched in Poultons reagent to reveal the macroscopic flow. The remaining half was further sectioned, removing thin layers ($500\mu\text{m}$) from the meridian plane, using a wafering wheel. These foils were then ground to $300\mu\text{m}$

using 1200 silicon carbide paper and 3mm discs punched from the foils at locations of interest indicated from the etched half of the billet. These discs were subsequently jet thinned as described in Chapter 4 and examined on the TEM.

The macroscopic flow patterns from these trials can be seen in Plate (6.8.). All the trials had common process conditions of strain-rate $(\dot{\epsilon})=33s^{-1}$ and an extrusion ratio of 80:1. The direct extrusion trials were processed at 550°C, 450°C, and 350°C and can be seen in Plate (6.8.) a, b and c respectively. The indirect extrusion trial was processed at 450°C (Plate (6.8.d)). As the initial billet temperature is increased, in the directly extruded partials, there is a corresponding increase in the deforming volume. There is also an increase in the angle made by the DMZ. These features are in keeping with the results from the 'gridded billet' trials. The intense shear regions in the high temperature partial have been somewhat 'masked' by static recrystallisation which occurred between trial cessation and quenching. This indicates that the temperature in the deformation zones is higher than the remainder of the billet. The local heat sources within the deforming billet is to be expected and emphasises the need to incorporate such detail into the extrusion thermal model⁽¹⁴⁾. The flow pattern produced in the indirect mode indicates that deformation is limited to a very small portion of the billet, adjacent to the die throat, confirming the more homogenous nature of the indirect extrusion process.

In view of the variance in flow patterns shown in Plate (6.8.), a more detailed examination of the developing structure was made using the TEM. In Plate (6.9.), micrographs from the 350°C direct extrusion

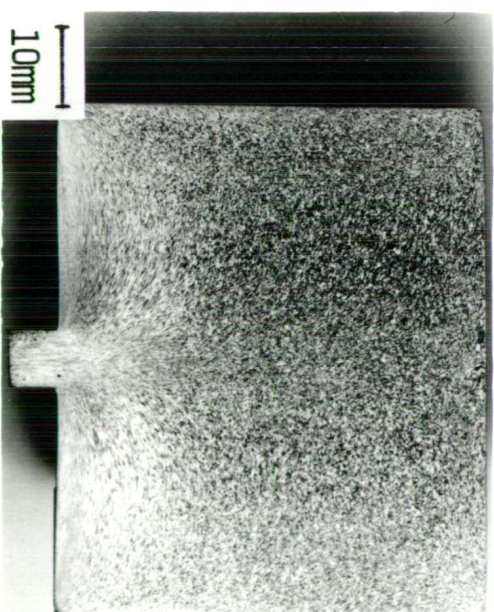
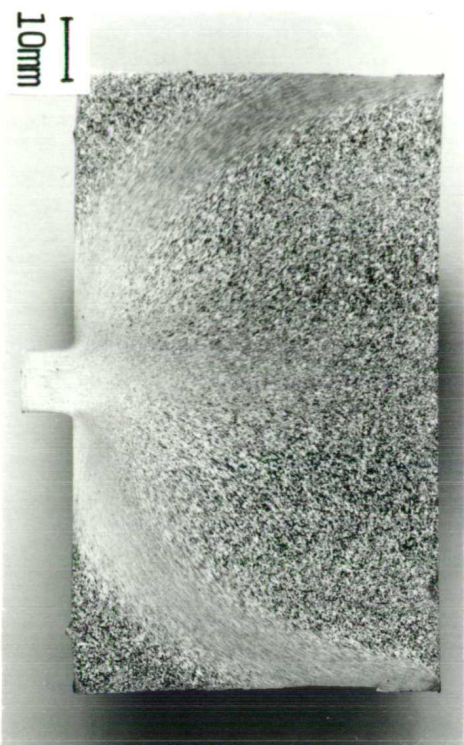
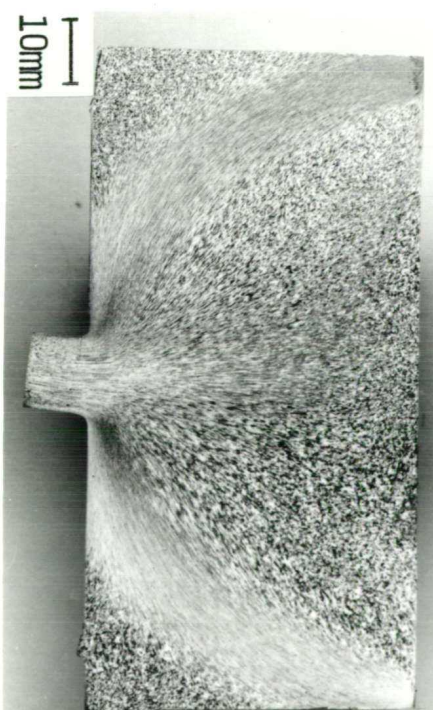
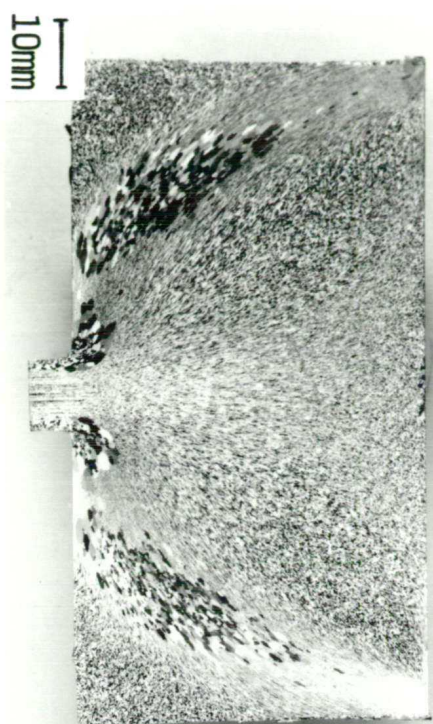
(a) Low Z Partial Extrusion (direct)

(b) Medium Z Partial Extrusion (direct)

(c) High Z Partial Extrusion (direct)

(d) Medium Z Partial Extrusion (indirect)

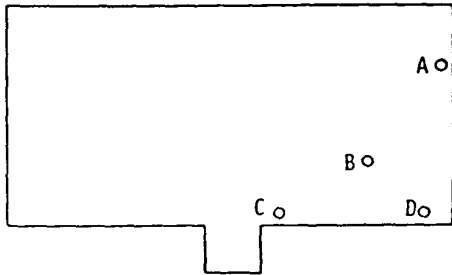
Plate 6.8. Quasi-static extrusion flow patterns



trial can be seen. Also indicated are the locations of specimen extraction. At position 'A' (Plate (6.9.a)), which is located near the rear of the billet, within the shear zone, the sub-structure is already relatively well defined and equiaxed. There is a fairly low internal dislocation density and the larger second phase particles are aligned in the extrusion direction. At position 'B' (Plate (6.9.b)) midway through the deformation zone, the sub-grain structure is similar to that at 'A', indicating that the sub-structure may already be fully developed between 'A' and 'B'. The sub-structure is quite obviously changing by a combination of dynamic recovery and repolygonisation.

The increased strain suffered by material at 'B' is reflected in the further alignment of second phase precipitates. At position 'C' (Plate (6.9.c)), adjacent to the die mouth, the sub-structure is well developed. The sub-grains exhibit 'banding' which is thought to be due to containment in the extrusion direction. The high dislocation density apparent at this location is consistent with that required to cause static recrystallisation at the annulus of the extrudate. In the DMZ, (position 'D'), the metal has received compressive deformation resulting in a high dislocation density (Plate (6.9.d)). As the stagnant metal in this region is at a temperature within the range for maximum rejection of Mg_2Si , precipitation occurs. The needle shaped β' precipitates can be seen in Plate (6.9.d). As well as the needle shaped precipitates, there is a small quantity of round particles which are, in fact, β' needles laying in the foil axis direction, being viewed end on.

An increase in initial billet temperature to $450^\circ C$ is accompanied by a broader deformation zone, containing sub-grains of a slightly larger



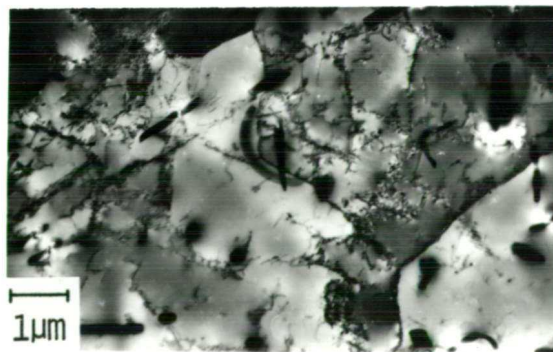
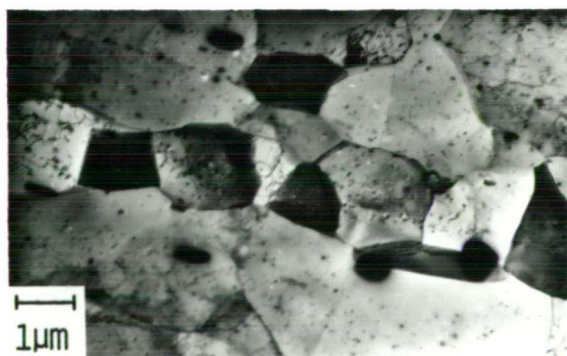
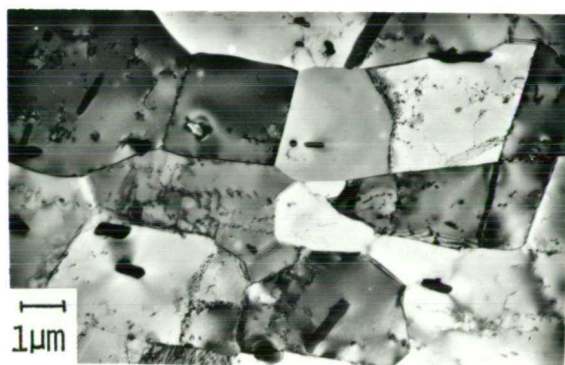
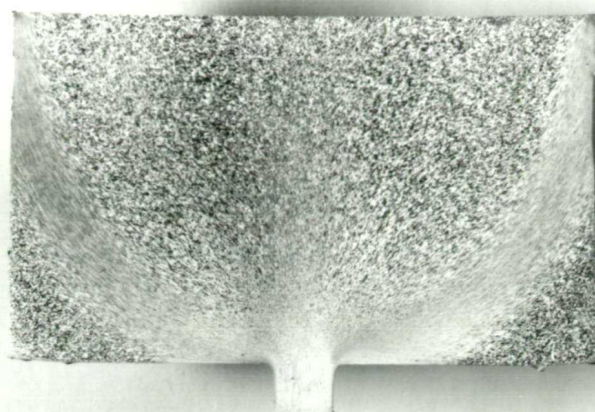
(a) Location A

(b) Location B

(c) Location C

(d) Location D

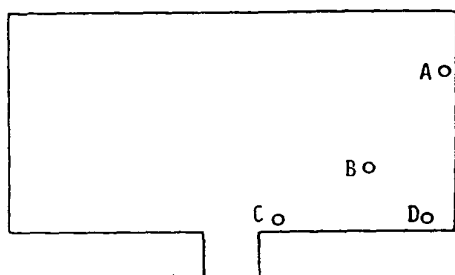
Plate 6.9. Structural variation in deformation zone (350°)



size than the previous condition. At position 'A', shown in Plate (6.10.a), the sub-structure adjacent to the start of the deformation zone can be seen. The sub-grains are quite irregular and contain a relatively high dislocation density. This indicates that conditions for steady state repolygonisation have not yet been reached. The low strain imposed at this location is reflected in the non-alignment of second phase precipitates. At location 'B' within the deformation zone the sub-structure has developed considerably (Plate (6.10.b)). The sub-grains are more regular and equiaxed but continue to exhibit a high dislocation density. The alignment of the intermetallic iron phase is indicative of the high proportion of strain received by the metal at this location. Small needle shaped Mg_2Si precipitates can also be observed within the microstructure shown in Plate (6.10.b). These probably formed before the billet was quenched after the trial was stopped.

Once the metal has reached the die throat, at location 'C', the sub-grains take an equiaxed form with sharply defined sub-boundaries (Plate (6.10.c)). The sub-grain size is similar to that at 'B' but has greater internal order. This results from the greater thermal energy at this location promoting greater cross-slip and climb of the continually rearranging dislocations. The metal within the DMZ (location 'D') is again heavily dislocated. However, due to greater dislocation mobility sub-boundaries have formed populated with a high density of dislocations.

At the highest extrusion temperature employed from the direct extrusion trials (550°C) structural development is rapid. The increased thermal energy allows greater dislocation activity resulting in an almost



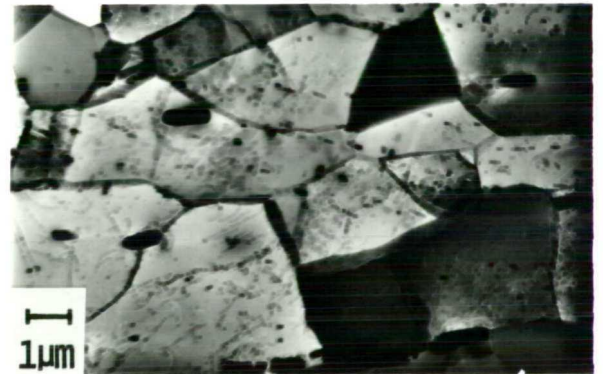
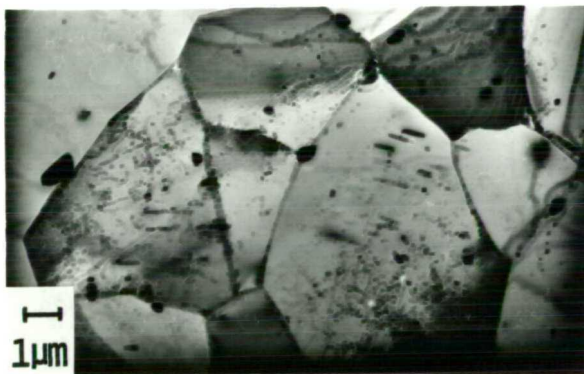
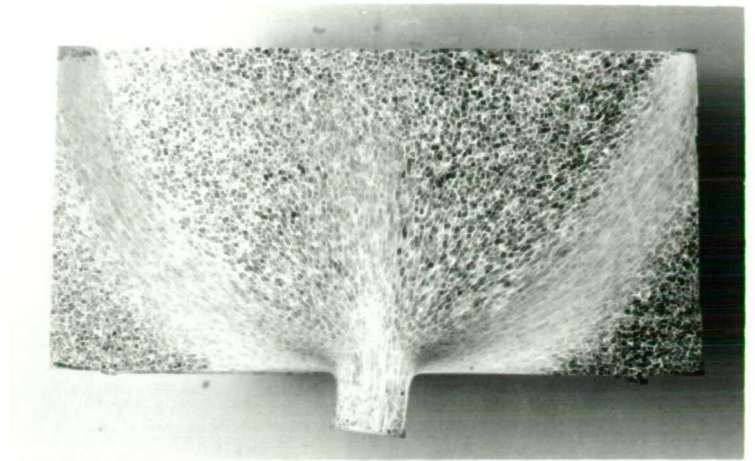
(a) Location A

(b) Location B

(c) Location C

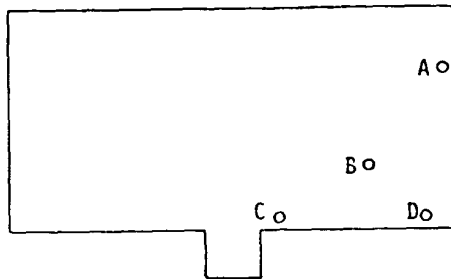
(d) Location D

Plate 6.10. Structural variations in deformation zone (450°C)



constant sub-grain size throughout the deformation zone as a consequence of repolygonisation. (See Plates (6.11.a-c). The small precipitates evident in Plate (6.11.a,b,d) are equilibrium (β) Mg_2Si .

The indirect trial, performed at 450°C , demonstrated that material in the deformation zone flows parallel to the die face (Plate (6.8.d)). Bearing this in mind, specimens were taken from different positions to those in the direct trials. These are shown in Plate (6.12.). At location 'A' the structure at the edge of the deformation zone near the billet container interface can be seen (Plate (6.12.a)). There is evidence of sub-grain formation though this is not complete. The structure at this location is comparable with location 'A' in the direct mode shown in Plate (6.10.a). Midway through the deformation zone at location 'B', the sub-grains are well developed and there is an alignment of AlFeSi parallel to the die face (Plate (6.12.b)). Upon reaching position 'C', adjacent to the die throat, the sub-structure retains a similar appearance to that at 'B'. The structure at the die throat bears strong similarities to the equivalent position in the direct mode (Plate (6.10.c)). On the outskirts of the deformation zone (position 'D') there is a large variance in the sub-grain size (Plate 6.12.d)). This is a result of the strain gradient present between the small deformation zone and the relatively undeformed metal forming the bulk of the extrusion billet towards the rear. In this rear portion (position 'E') the structure is one of original homogenised grains with high internal dislocation density produced during billet upsetting (Plate (6.12.e)). A more global representation of structural development for both modes of extrusion at 450°C can be seen in the optical micrographs in Plates (6.13. and 6.14.). For the direct



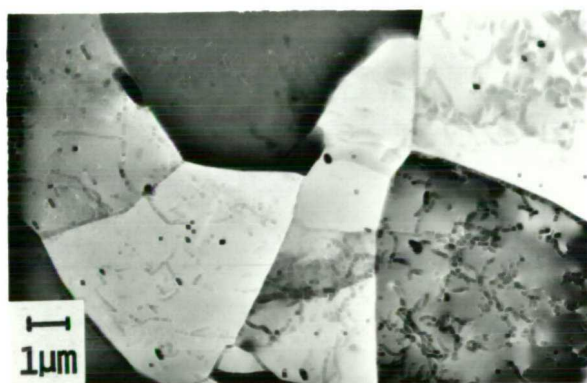
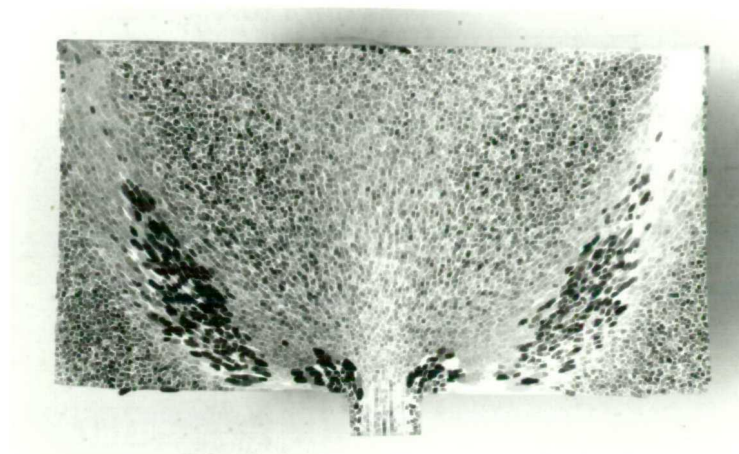
(a) Location A

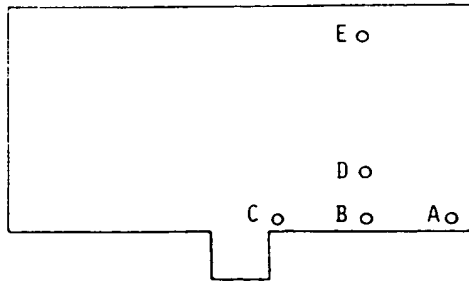
(b) Location B

(c) Location C

(d) Location D

Plate 6.11. Structural variations in deformation zone (550°C)





(a) Location A

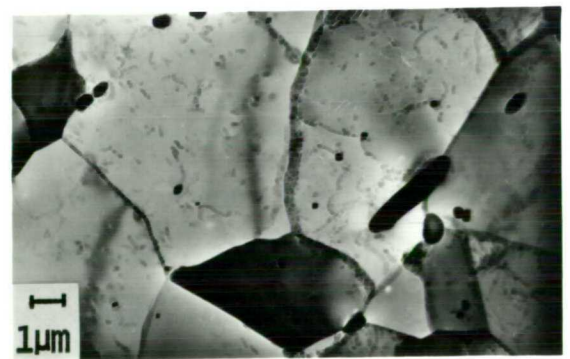
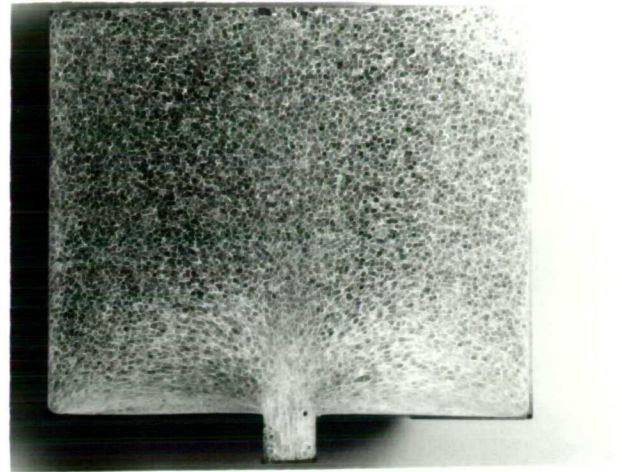
(b) Location B

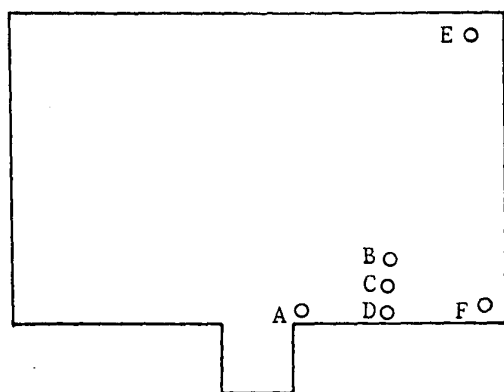
(c) Location C

(d) Location D

(e) Location E

Plate 6.12. Structural variations in deformation zone (indirect(450°C))





(a) Location A

(b) Location B

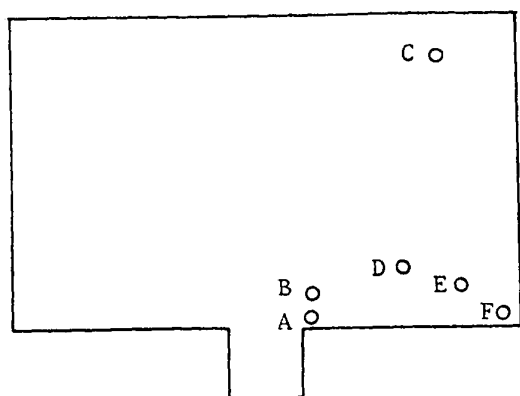
(c) Location C

(d) Location D

(e) Location E

(f) Location F

Plate 6.13. Structural variations in deformation zone (optical)



(a) Location A

(b) Location B

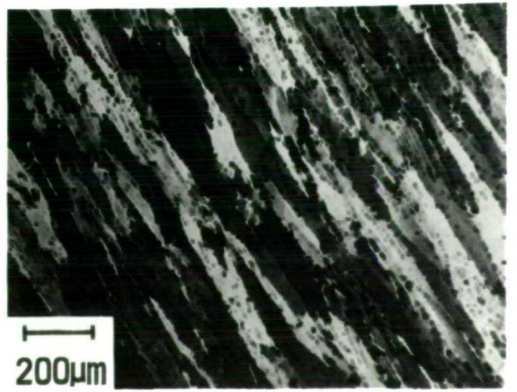
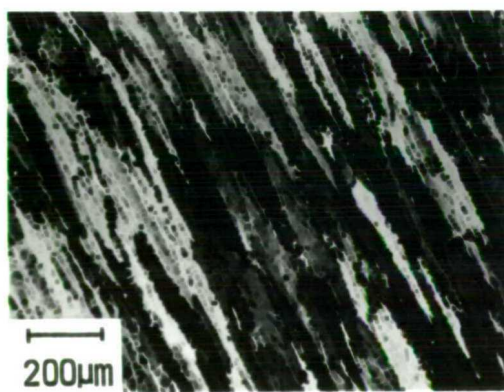
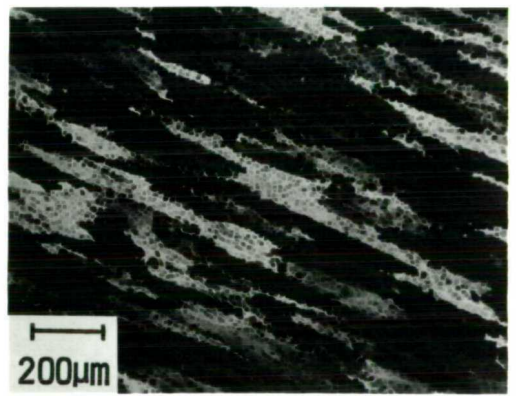
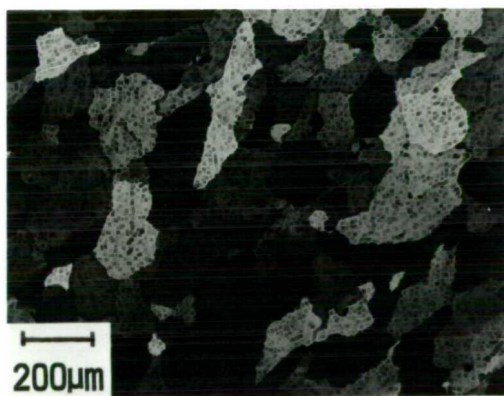
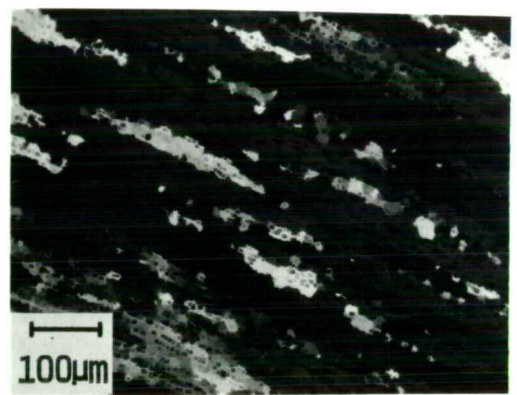
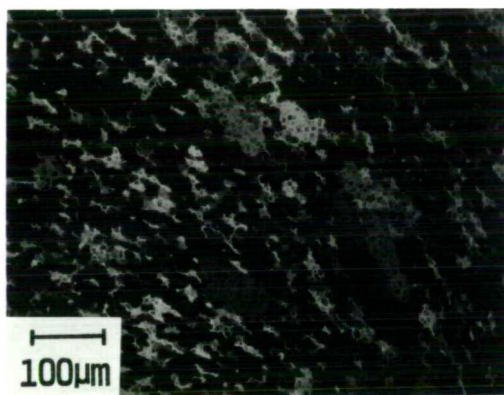
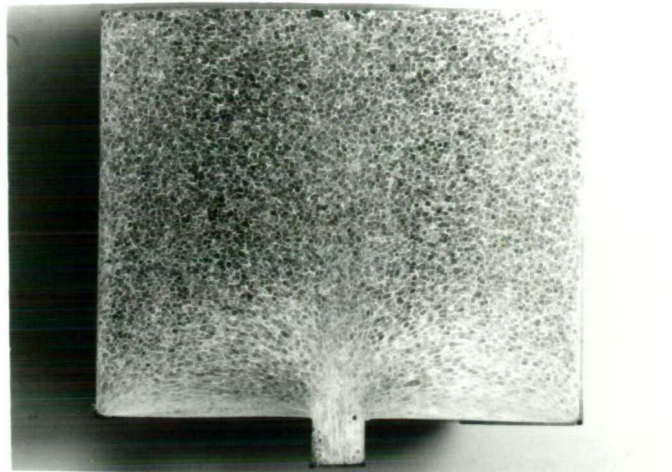
(c) Location C

(d) Location D

(e) Location E

(f) Location F

Plate 6.14. Structural variations in deformation zone (indirect(optical))



extrusion mode (Plate 6.13.) the relatively low deformation imposed on metal outside the deformation zone is quite clear, that is, at positions 'B' and 'E' shown in Plates (6.13.b,e). At positions 'C' and 'D' within the deformation zone and at position 'A' adjacent to the die throat, the intense shearing has produced the elongated (recovered) structure seen in Plates (6.13.a,c,d). Within the DMZ at position 'F', no evidence of deformation is apparent, exhibiting features of the original homogenised grains (Plate (6.13.f)). In the indirect mode, the majority of the billet towards the rear possessed a structure characterised by Plate (6.14.c). Within the shearing regions of the deformation zone at positions 'B', 'D', 'E' and 'F' it is clearly demonstrated that shearing is not as intense as that in direct extrusion (Plate (6.13.c)).

6.3. Room Temperature Tensile Properties of Extrudates

The benefits of sub-structural strengthening on room temperature mechanical properties are well documented⁽⁵⁰⁾. However, its use to achieve better mechanical properties is of greatest practical value in non-heat treatable alloys. The AA6XXX series alloys are precipitation-hardening materials which achieve their best properties in the fully heat treated condition. A consequence of the heat treatment is that the sub-structural benefits are considerably reduced and are often completely removed.

The majority of AA6063 extrusion sections, available commercially, are in the -T6 condition. (Artificially aged to peak hardness.) Ageing takes place first by clustering of magnesium and silicon atoms on

certain preferred crystallographic planes and then by atomic rearrangement and agglomeration to form coherent and partially coherent precipitates of Mg_2Si , with accompanying lattice strain, which produces the hardening. The maximum strain, and thus hardening, are probably obtained when the clusters have reached their greatest size and are on the point of losing coherency with the matrix⁽¹³²⁾.

In the present study, the room temperature tensile strength for extrudates produced in the 350°C-550°C temperature range, were obtained for the -O, -T1, -T4 and T6 conditions. The heat treatments pertaining to each temper is given in Chapter 4. Standard No.14 Hounsfield tensile specimens were machined from the extrudate at a position 0.5 of the ram stroke which was within steady state extrusion. The results from these trials, for a crosshead speed of $1mm.min^{-1}$, are shown in Table (6.1.) and are plotted in Figure (6.3.) together with the results quoted by the Aluminium Association.

The samples in the -T6 condition show good properties for all extrusion temperatures. The slight drop in the UTS value at 350°C is because of the high rejection of Mg_2Si at this temperature during extrusion, thereby resulting in coarse precipitates which may not totally dissolve during solutioning.

The most interesting features of the tensile results are found in comparing the -T1 and -T4 tempers. These tempers are essentially the same, apart from their solutionising treatment; -T4 is solutionised after hot deformation, whilst -T1 is solutionised as a result of hot deformation. It is clear that subsequent to natural ageing the -T1 samples have an increment of strength over the -T4 samples. This is

		outs	oy	% Elong
350 (168d)	-O -T1 -T4 -T6	11 189 181 246	51 109 86 218	54 38 49 29
400 (172d)	-O -T1 -T4 -T6	- 176 182 254	- 100 87 215	- 40 46 30
450 (171d)	-O -T1 -T4 -T6	84 177 179 256	47 90 85 216	51 43 44 32
500 (170d)	-O -T1 -T4 -T6	84 184 177 256	48 86 86 218	51 43 43 32
550 (169d)	-O -T1 -T4 -T6	85 196 178 258	46 99 85 218	50 46 44 31

Table 6.1. Tensile Properties for AA6063

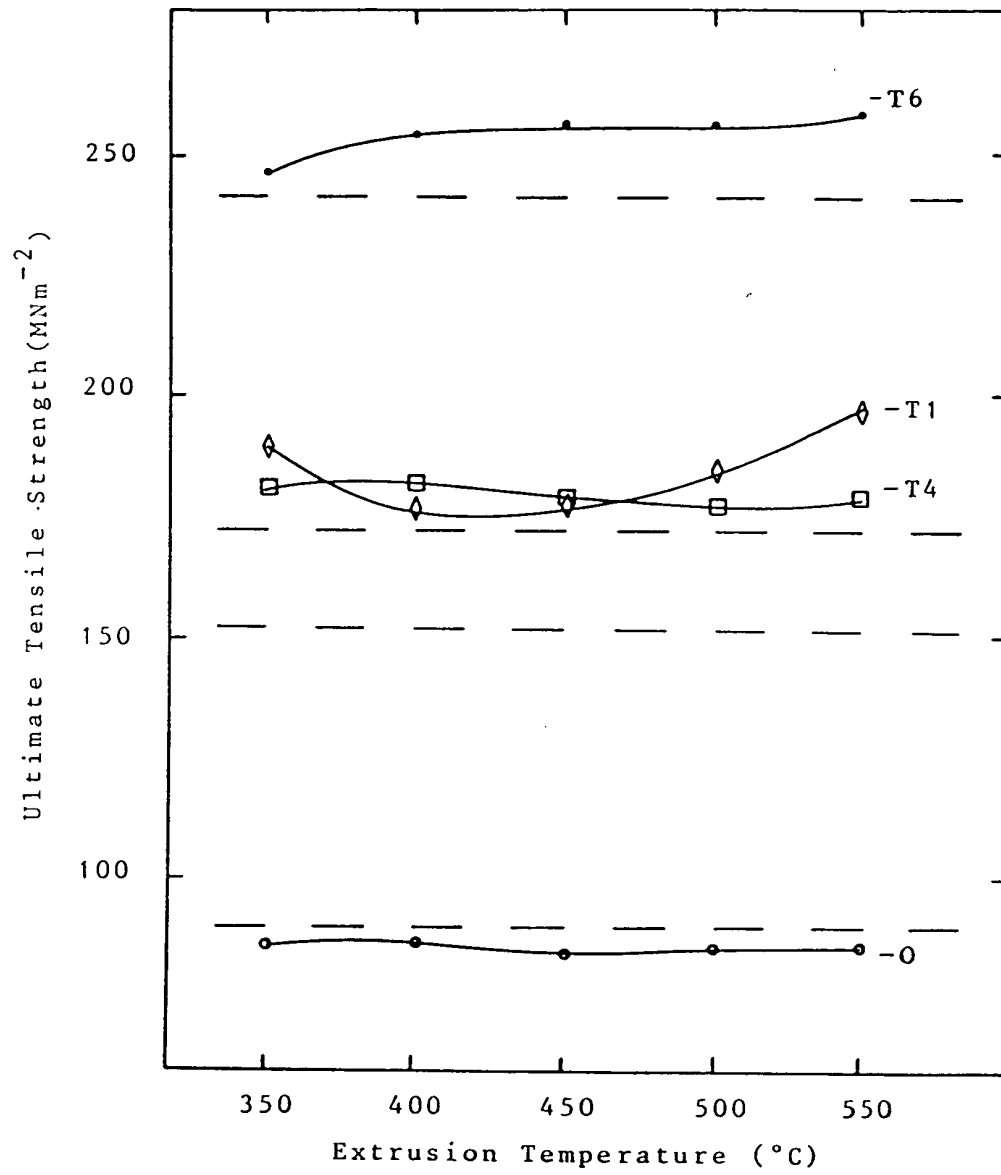


Figure 6.3. (a) U.T.S.

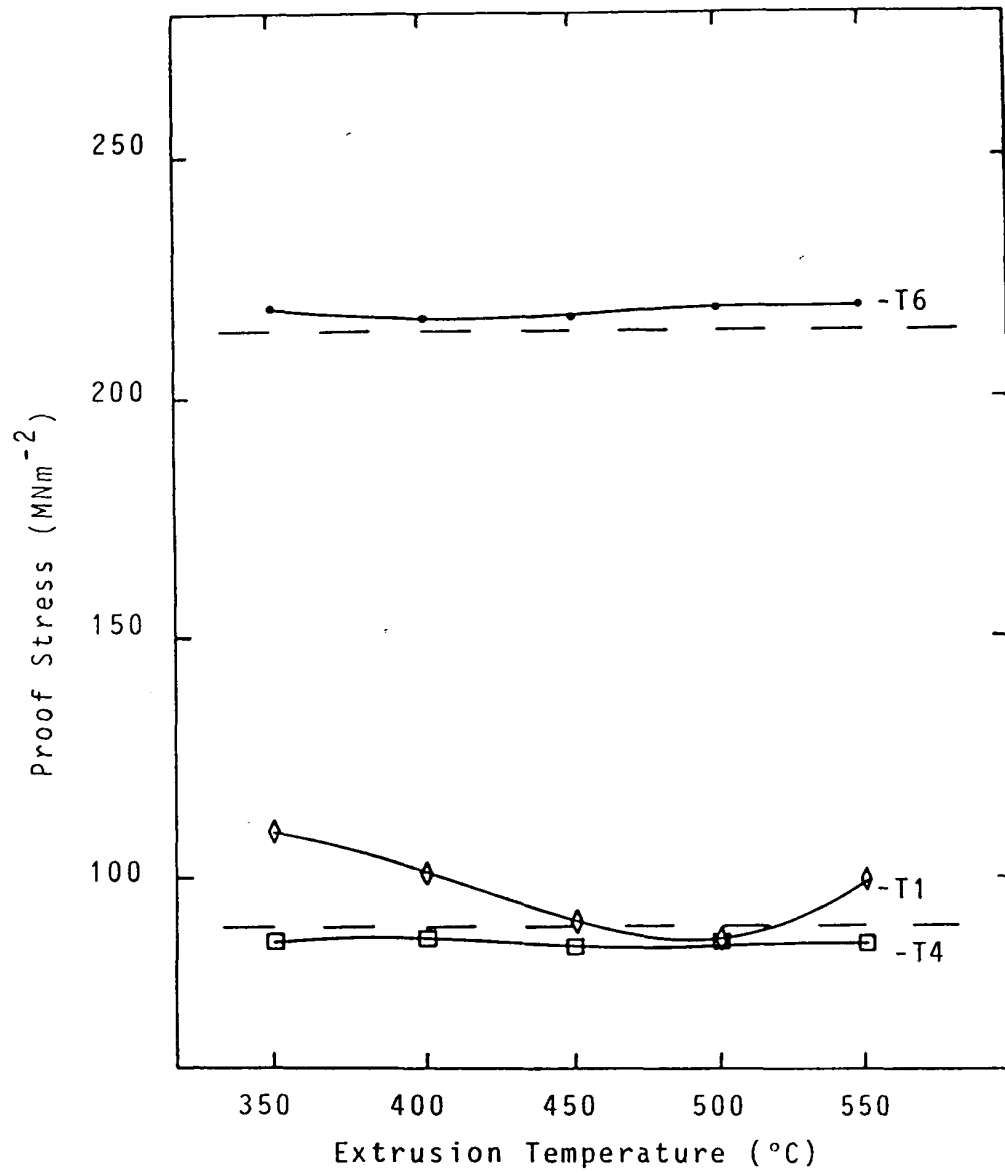


Figure 6.3. (b) Proof Stress

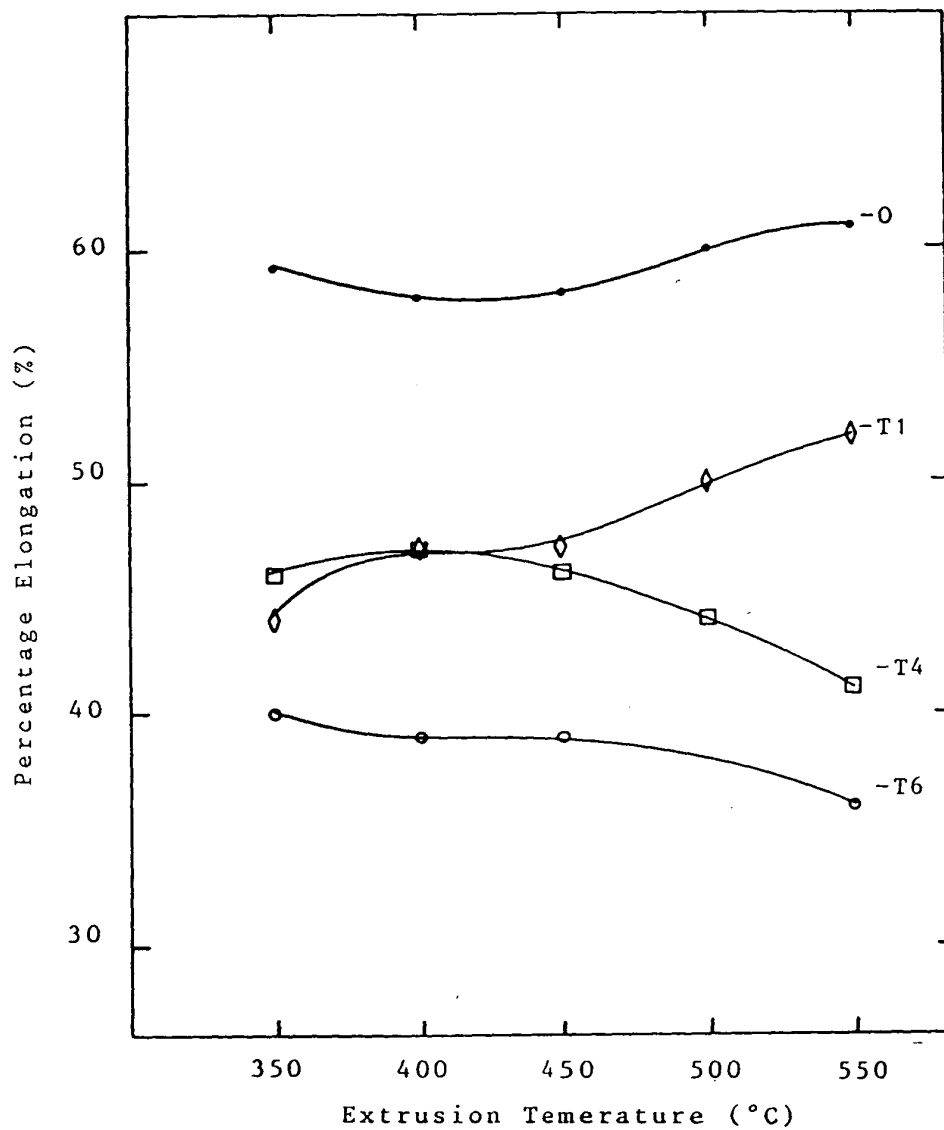


Figure 6.3. (c) Percentage Elongation

because the solutionising treatment performed on the -T4 samples eliminated the hot worked structure by static recrystallisation, and it is well-known⁽⁵⁰⁾ that the sub-structure produced by hot working improves tensile properties. With suitable quench conditions, the -T1 samples retain the sub-grain structure produced during extrusion. This explains the steady increase in both yield strength and UTS for the samples extruded at 450°C to 350°C; it was shown earlier that a decrease in extrusion temperature produced a smaller sub-grain size and there is considerable documented evidence⁽⁵⁰⁾ that a smaller sub-grain size improves tensile properties.

There is also evident, in Figure (6.3.) an increase in tensile properties for the -T1 condition at an extrusion temperature of 550°C, which quite clearly can not be attributed to the substructural strengthening. In this event the strength increment is from increased solution strengthening as the solid solubility increases with temperature. In addition, the high temperature extrusion conditions might be sufficient to cause further β - α iron phase transformation thereby releasing more Si making it available for hardening in conjunction with Mg.

The effect of extrusion conditions on the -T1 tensile properties therefore depends on whether the extrudates are a function of sub-structure strengthening (at low temperature or high Z) or the amount of solute in solution (a high temperature or low Z). Whereas the -T4 and -T6 conditions depend entirely on the solution treatment and ageing process, due to the loss of the hot worked structure.

6.4. Extrusion Limit Diagrams

There are two limitations to processing a satisfactory product by extrusion. One limiting factor lies in the inherent stiffness of alloys, even at the elevated temperatures, which may result in the pressure required exceeding the press capacity. The other major factor limiting the process parameters is the maximum temperature which may be tolerated before the onset of deleterious surface defects.

The graphical representation of extrusion limits within which products with an adequate surface finish could be produced for a specific press capacity, was originally proposed by Hurst and Ursell⁽¹²⁶⁾. This form of diagram is shown in Figure (6.4) for AA6063, for both direct and in direct modes of extrusion. The limiting pressure lines described by the maximum extrusion ratios and minimum billet temperatures have been established using the empirical pressure equation in Section 5.3.5. for a press capacity of 1130MNm^{-2} . The lines representing surface failure are described by the expressions derived in Section 7.3.1.

Also shown in Figure (6.4.) are the process limits for alloys AA2014⁽¹¹⁾ and AA5456⁽¹⁴⁾. It is quite clear that AA6063 has a much higher extrudability than the more heavily alloyed AA2014 and AA5456 due to the retarding effect on dislocation activity and the formation of low melting point phases. In fact, the AA6063 limit lines are reminiscent of the diagram given for pure aluminium by Sheppard and Raybould⁽¹³³⁾.

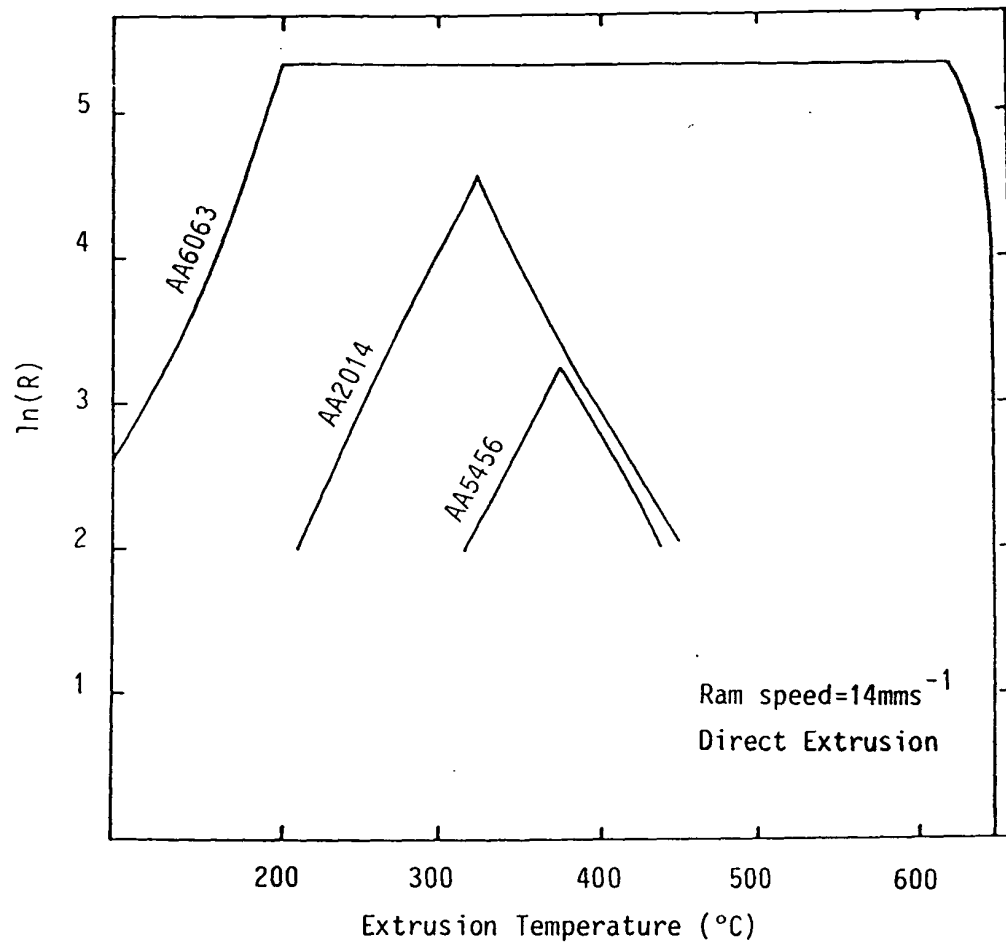


Figure 6.4. AA6063 Limit Diagram (with AA2014 and AA5456 as a Comparison)

The limiting pressure line on the basic limit diagram demonstrates that as the initial billet temperature increases, the flow stress decreases enabling higher extrusion ratios. This continues until a critical temperature and extrusion ratio are reached, where an increase in temperature results in a decrease in the possible extrusion ratio due to the increased likelihood of surface cracking as a result of the temperature rise during extrusion.

The usefulness of limit diagrams has been improved since the first plot of Hirst and Ursell. Numerous workers have shown how the diagrams can be used to display information on metallurgical structure and hence provide the possibility of metallurgical process control. It was shown in Section 6.2.1.2. that a relationship between the sub-grain size and the temperature compensated strain rate could be formulated. This quantification allows these results to be represented on the extrusion limit diagram. In Figure(6.5.) lines of constant sub-grain size are plotted, demonstrating that a particular sub-grain size may be obtained at numerous temperature-extrusion ratio pairs. Sheppard⁽¹²⁴⁾ has shown that these sub-structure lines have the added advantage of representing room temperature mechanical properties, as these are related to the sub-grain size (and hence 'Z'). However, the results in Section 6.3. showed that for heat-treatable alloys such as AA6063, the relationship between process conditions (and subsequent structure) and the mechanical properties is not so simple.

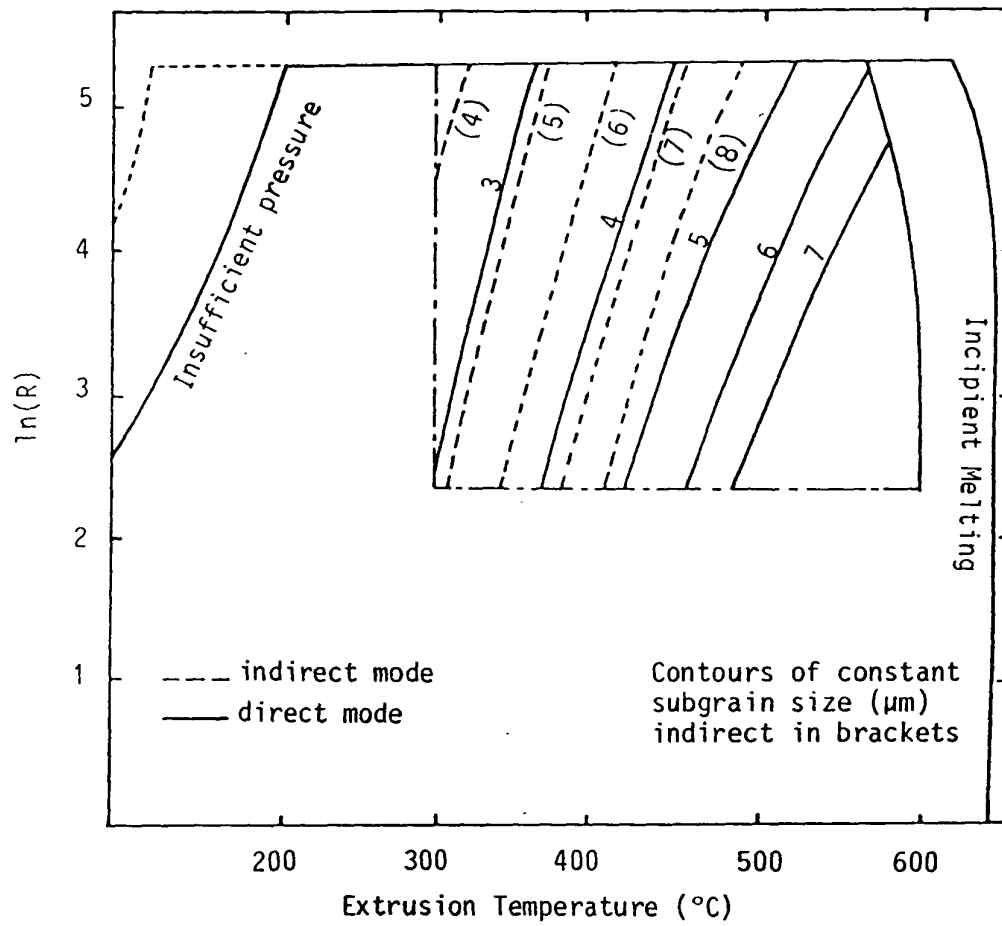


Figure 6.5. AA6063. Limit Diagram with Structural Information

Conclusions

- (1) The heterogeneities in the as-cast AA6063 billet require an homogenisation treatment to be performed in order to improve extrudability and surface generation.
- (2) The most important transformation during homogenisation is β to α iron phase.
- (3) Homogenisation leaves the magnesium and silicon in a suitable form for rapid dissolution during hot deformation; that is β' form.
- (4) Three types of structure are produced during hot extrusion of AA6063. Duplex, elongated recrystallised and recrystallised. The differing structures were a consequence of the process conditions and extrudate cooling practices.
- (5) The operation of dynamic recovery during deformation produces a sub-grain structure. The operation of static recrystallisation immediately after extrusion often eliminates this sub-structure.
- (6) The sub-grain size, 'd' produced during torsion and both modes of extrusion may be related to the process conditions by an equation of the form:

$$d^{-1} = a + b \cdot \ln Z$$

- (7) The quasi-static flow pattern and the development of structure within the extrusion billet is determined by the process conditions and mode of extrusion.
- (8) The morphology of the structure and sub-structure produced in torsion and both modes of extrusion are essentially the same. This allows the torsion test to be used as a model deformation process.
- (9) Room temperature tensile properties are generally determined by solution treatment and ageing. An increment of strength can be achieved when a retention of sub-structure is possible.
- (10) The extrusion limit diagram can be used to represent process limits for a given press capacity. The limit diagram framework can also be used to display structural information

CHAPTER SEVEN

CHAPTER SEVEN

Material Flow and Surface Generation During Extrusion

7.1. Introduction to Surface Generation and Surface Defects Encountered

During the extrusion of aluminium alloys, especially those for high strength aerospace application, the attainment of good mechanical properties is one of prime importance. However, these extrudates are often rendered useless by the onset of the deleterious surface defect, 'hot shortness' (sometimes affectionately known as 'fir tree defect' from its appearance). Thus it is often the surface generation that governs productivity. In the common low strength alloys, such as AA6063, it is the cosmetic appearance which is of prime concern. Although these alloys have a very high extrudability compared with the low strength series of alloys, it is surface defect formation that again limits production. In these alloys it is the occurrence of 'pick-up' and 'die lines' that is most common. These surface defects can cause severe degradation of the extrusion product's aesthetic quality, thereby necessitating additional post-extrusion surface treatments.

In this chapter, the commonly occurring surface defects will be discussed, with special attention given to pick-up and die lines. To study the formation mechanisms of the defects, the material flow and

surface generation mechanisms, operative during extrusion, must first be examined.

7.2. Surface Generation

In order to achieve the best possible surface quality from extrusion, the surface of the extrudate must be generated from within the billet. It is therefore common practice in hot extrusion of aluminium alloys to process without lubricant and through flat faced (180°C) dies. When direct extrusion is performed in this way, sticking friction normally exists at the interface between the tooling and the billet material. This tends to retard the flow of material at the billet surface, promoting the continuous formation of a 'virgin' surface shearing from within the billet. (Flow pattern 'B'.) However, when extrusion is performed in the indirect mode, the nature of the metal flow results in the billet surface forming part of the extrudate surface. (Flow pattern 'S'.)

The major portion of the present research was concerned with the direct mode of extrusion. The features of the metal flow during extrusion performed in this mode were studied using the techniques; macro (structure) etching, and pinned billets.

7.2.1. Material Flow Features

Macro (Structure) Etching

The quasi-static flow of metal during unlubricated extrusion, through a flat face die, is shown in Plate (7.1.a) for a strip section. The sticking friction present at the container wall/billet interface has promoted the formation of a dead metal zone at the front of the billet, adjacent to the die throat. This volume of stagnant metal extends from the die throat, back towards the container wall, at a typical angle of around 45° for AA6063 at hot working temperatures. Outlining this zone is a region of intense shear called the deformation zone. This region is labelled in Plate (7.1.) and can be clearly distinguished in Plate (7.1.a). The zone is hemispherical and extends from the die throat to the container wall towards the rear of the billet. It is material from the deformation zone running along the dead metal zone boundary that forms the extrudate surface.

Also visible in Plate (7.1.a) is a central region of intense shear feeding into the bulk of the extrudate. This central deformation zone supplements the material feeding the body of the extrudate from within the main deformation zone.

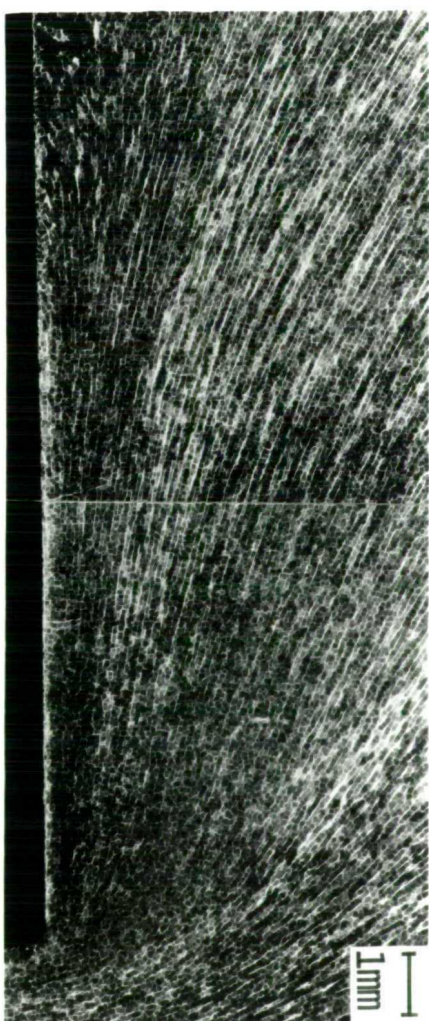
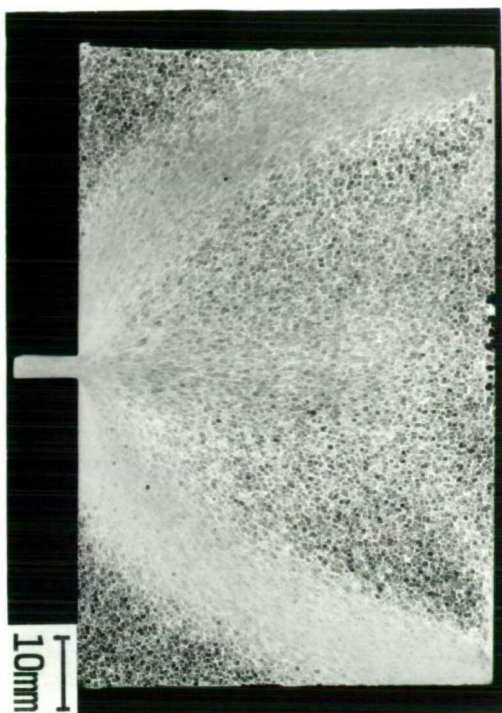
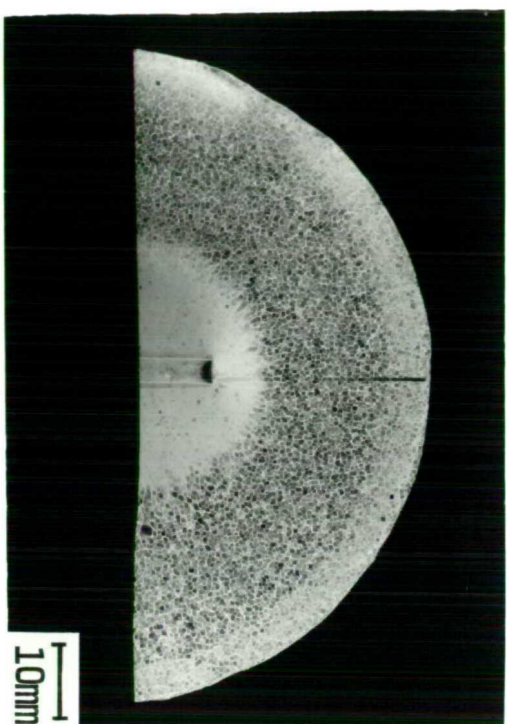
A third region of the extensive deformation can also be detected on the die face, adjacent to the die throat, feeding into the extrudate surface. This small volume region is shown clearly in Plate (7.1.b). The material residing in this volume is at a lower temperature than

(a) Quasi-static flow (meridian plane)

(b) Deformation zone on die face

(c) Quasi-static flow (billet face)

Plate 7.1. Features of Quasi-static Flow in Strip Extrusion.



the adjacent deformation zone due to the quenching effect of the die. (The die temperature at the commencement of extrusion was 100°C below that of the billet in all experimental trials.)

The bottom of the extrusion billet, shown in Plate (7.1.c), reveals that the deformation zone is almost concentric with the billet. Therefore the distance from the die throat, occupied by this deformation zone, is greater for the minor axis of the section. It is in this volume of metal that coarse iron-bearing precipitates collect, to be subsequently released into the die land/extrudate surface junction, thereby influencing the surface generated features.

The surface of the extrudate thus consists of material flowing from the region outlining the dead metal zone and a smaller quantity from the small deformation volume on the die face. The metal forming the dead metal zone/deformation zone boundary originates from the forward position of the extrusion billet.

7.2.2. Pinned Billets

Information regarding the surface formation mechanisms in extrusion can not be totally established by studying the structural flow alone. The most appropriate experimental technique to investigate the surface formation is by introducing volume elements into the billet so that they might be traced through the flow pattern. In this study, the introduction of 12 annealed copper pins into the extrusion billet allowed further knowledge to be gained. The arrangement of the pins can be seen in Figure (7.1.a). The billets were carefully aligned

prior to the extrusion such that the pins were positioned in line with the major and minor axis of the strip die. The prepared billets were then partially extruded at 450°C. The partially extruded billets were subsequently sectioned; one along the major axis, the other along the minor axis. After polishing and etching the cut faces, the material flow features were revealed. These can be seen in Plate(7.2.a,b).

The pins inserted at the rear of the billet have suffered relatively little deformation and can be seen flowing slowly into the bulk of the extrusion billet.

At the front end of the billet the radially inserted pins have deformed, creating the region of the deformation zone that is responsible for feeding the extrudate surface. In Plate (7.2.a.) traces of copper can be seen feeding along the dead metal zone boundary and into the extrudate surface. The smaller deformation zone on the die face, adjacent to the die throat, can also be distinguished, feeding concurrently into the extrudate surface.

The material flow along the steeper dead metal zone, of the major axis, is faster than that of the minor axis. This is because, in order to obtain a sound product, there must be the same net rate of material flow per unit length across any section, taken at the die throat. Thus there must be a radial flow of material in the deformation zone, such that the required net rate of material flow is achieved.

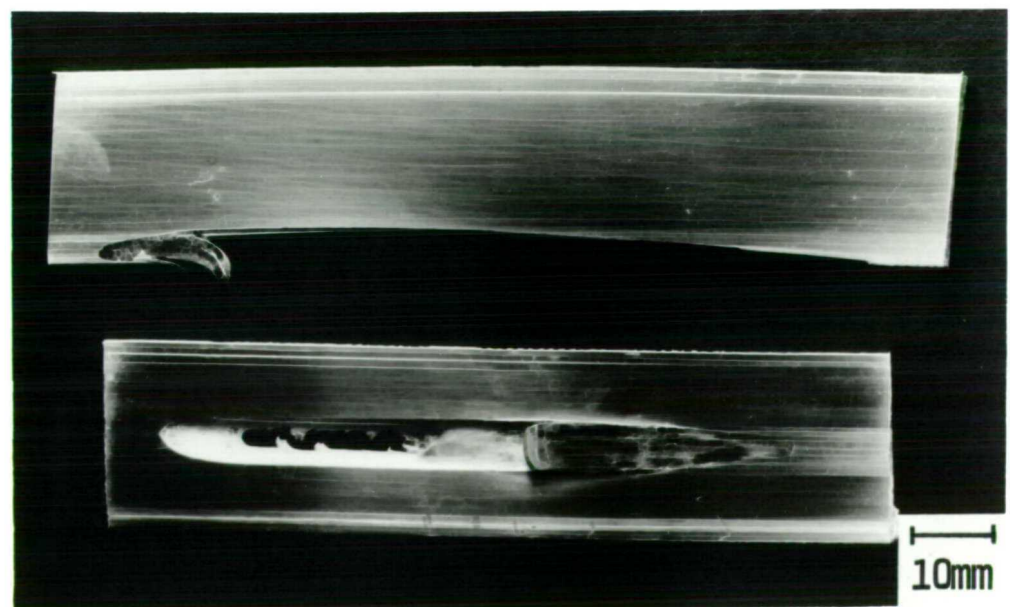
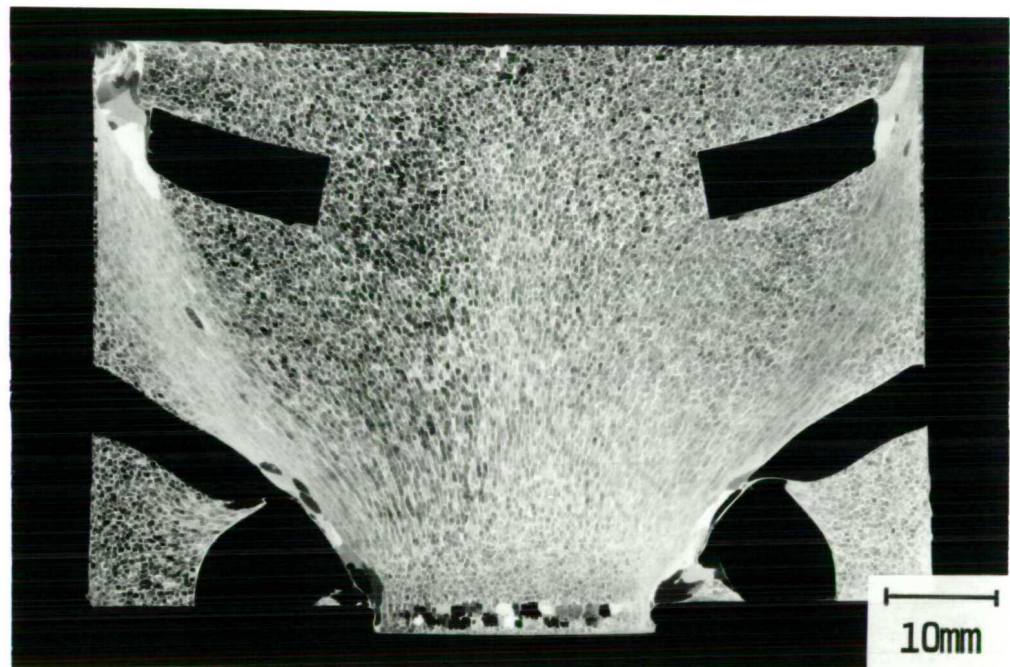
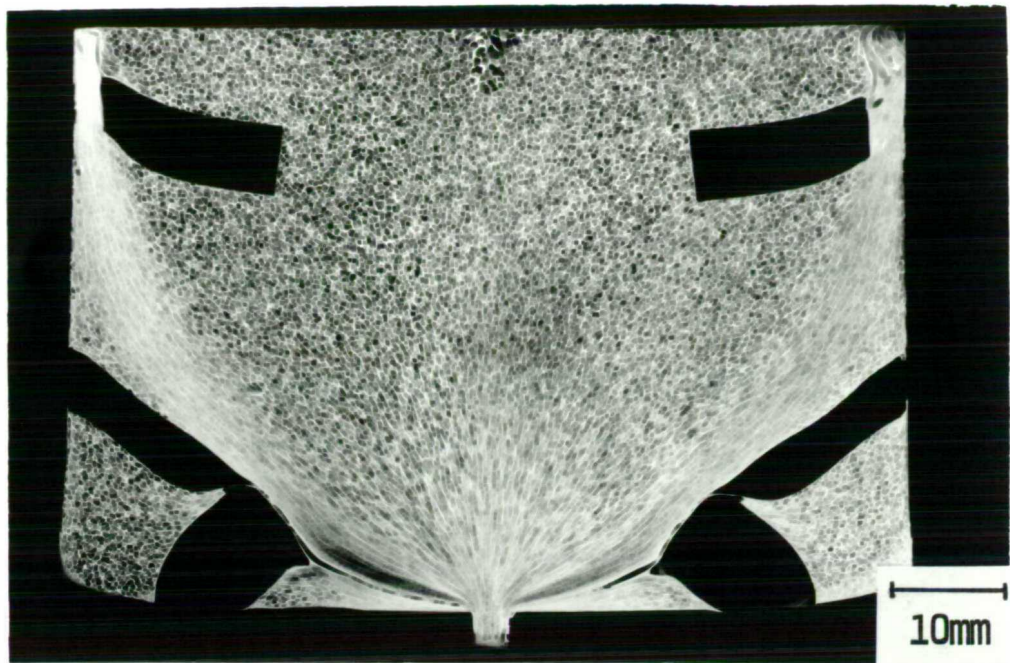
The radial pin located in the major axis is therefore the first to appear at the extrudate; in this case 1.3m from the front, whilst the

(a) Quasi-static flow (minor plane)

(b) Quasi-static flow (major plane)

(c) Appearance of pin at extrudate surface

Plate 7.2. Pinned Billet Flow Patterns.



radial pin flowing along the minor axis appears at 1.56m from the front (Plate (7.2.c.)). Once the pin becomes part of the steady state flow and appears at the extrudate surface, the element can be detected along the remainder of the extrudate surface, as a thin smear. This can be clearly distinguished on the wide face of the extrudate profile shown in Plate (7.2.c.).

These experiments provide further confirmation that for unlubricated extrusion through flat faced dies, the surface originates from the front

portion of the deformation zone by a continuous interaction and shearing of the dead metal zone boundary. The metal from this region is then supplemented by the small deformation volume located on the die face.

Because the volume elements introduced into the extrusion billet have different deformation and flow stress characteristics at hot working temperatures, the flow pattern is slightly distorted to that occurring in a uniform billet. This could be remedied by selecting a suitable combination of aluminium alloys with the same hot working characteristics but possess different etching characteristics. Also, the introduction of smaller volume elements would present less interference to homogenous flow. However, this would increase the difficulty in sectioning the billets to reveal the pins.

Despite the slight distortion of the flow pattern, the results obtained allow the various features of the quasi-static deformation to be indentified. These are shown schematically in Figure (7.1.b.).

7.3. Surface Defects

7.3.1. Hot Shortness

Hot shortness occurs when the temperature rise at the extrudate surface is sufficient to cause incipient melting of the low melting point phases and eutectics or when the material's flow stress is unable to withstand the frictional forces imposed at the die land. The defect appears as surface cracks inclined at approximately 45° towards the rear of the extrudate. This can be seen in Plate (7.3.b).

Incipient melting is more frequent in high strength alloys^(7,11,14) especially when attempts are made to extrude material in the as-cast condition. The alloy under investigation here has a low alloy content and the formation of any low melting eutectics were removed during homogenisation. Therefore incipient melting only occurred when the extrudate temperature exceeded the matrix melting point of 643°C . It was rare that this temperature was exceeded apart from intentional trials to study this phenomena.

Incipient melting was always preceded to some extent by 'speed cracking'. This type of surface cracking is caused by a reduction in flow stress at the higher temperatures at this location. The occurrence of speed cracking becomes more likely as the strain-rate (or exit velocity) is increased; hence the term 'speed' cracking. The reason for this is explained by considering the temperature rise at the extrudate surface, as a result of the friction between the metal and the die lands. (Chapter 3.) This considerable heat source has a decisive influence on the extrusion speed and of course the initial

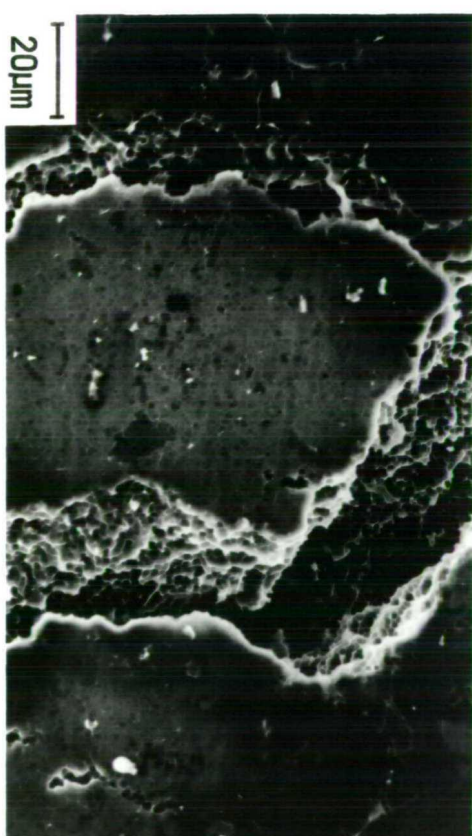
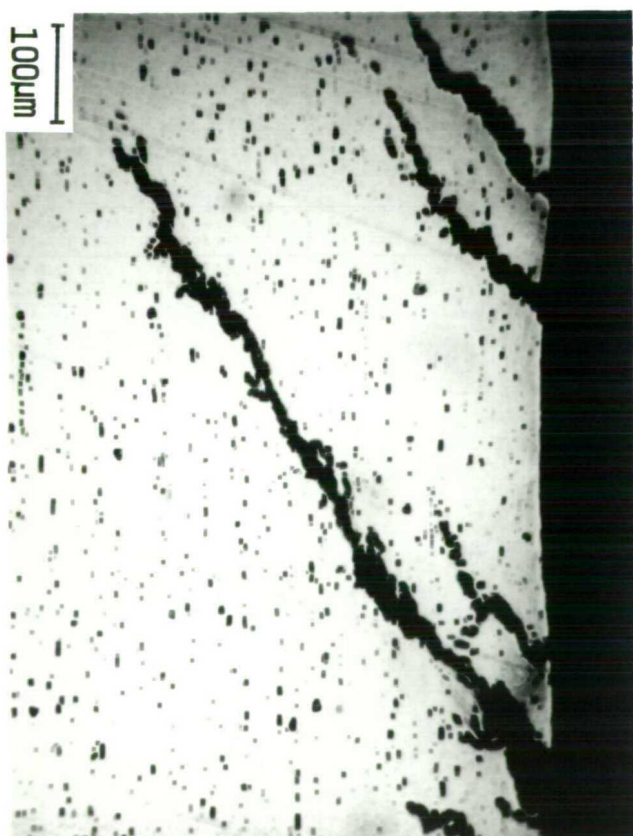
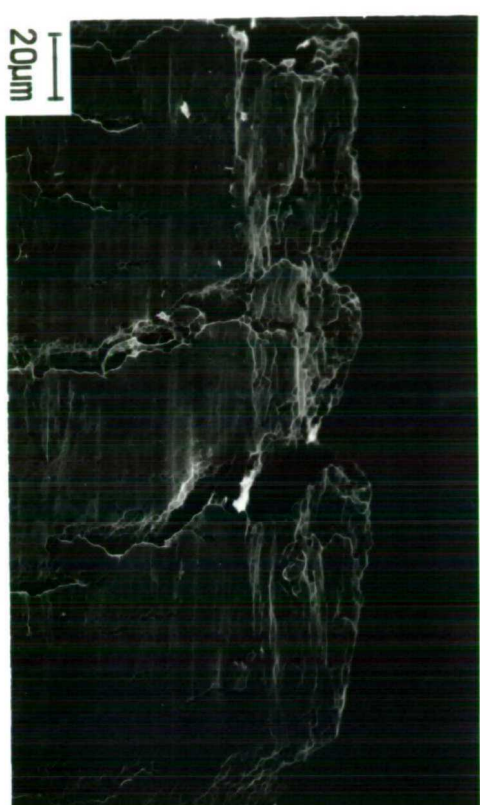
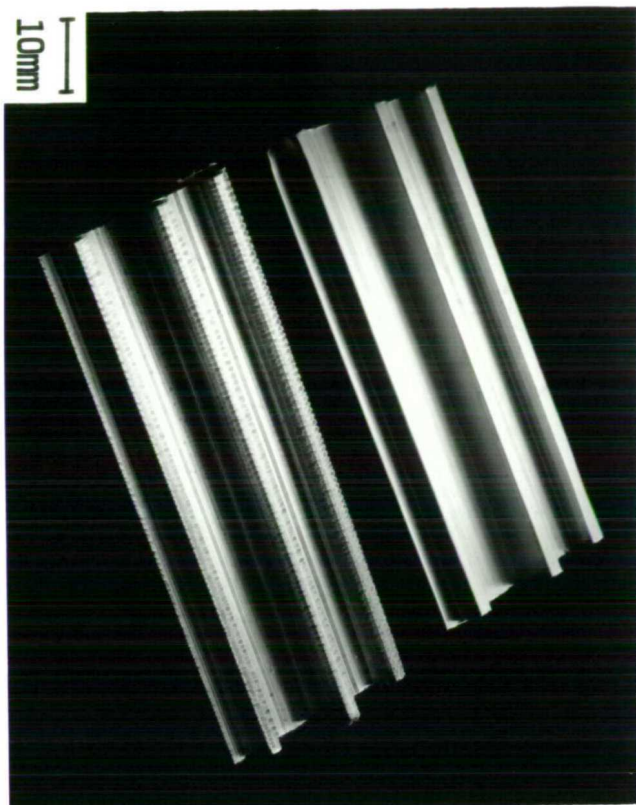
(a) Ribbed Section Extrusion

(b) Hot Shortness in AA6063 (optical)

(c) Surface cracking at corner location

(d) Network of incipiently melted material

Plate 7.3. Hot Shortness in AA6063.



billet temperature. It is quite common in extrusion, especially in AA6XXX series alloys, to employ high exit speeds. Therefore the brief contact time between extrudate and die land means that the temperature rise is concentrated in a thin surface layer. As the exit speed is increased the contact time is reduced thereby reducing the penetration depth. This feature is demonstrated in Figure (7.2.) using the equation developed in Chapter 3.

It must be remembered that this source of heat input is additional to that caused by total billet deformation⁽¹⁴⁾, which itself produces a temperature gradient across the extrudate; the severe deformation of the metal on passing the dead metal zone results in a higher temperature at the extrudate surface.

The hot shortness phenomenon is enhanced by the sudden loss of ductility, as the billet material moves into the die throat, from the release of the hydrostatic component of stress. This explains the commonly observed inclination of the surface cracks at approximately 45° towards the rear of the extrudate.

The danger of hot shortness increases when extruding complex sections especially when these include sharp corners. At such locations the propensity for crack formation is enhanced by two factors. The first is that the matrix at a corner point will receive almost double the frictional force, exerted by the perpendicular die lands. The second reason is that the volume of material receives a heat input from the two perpendicular die lands. Therefore the combination of increased temperature (and therefore reduced flow stress) and increased frictional forces tending to tear the matrix, the cracking will

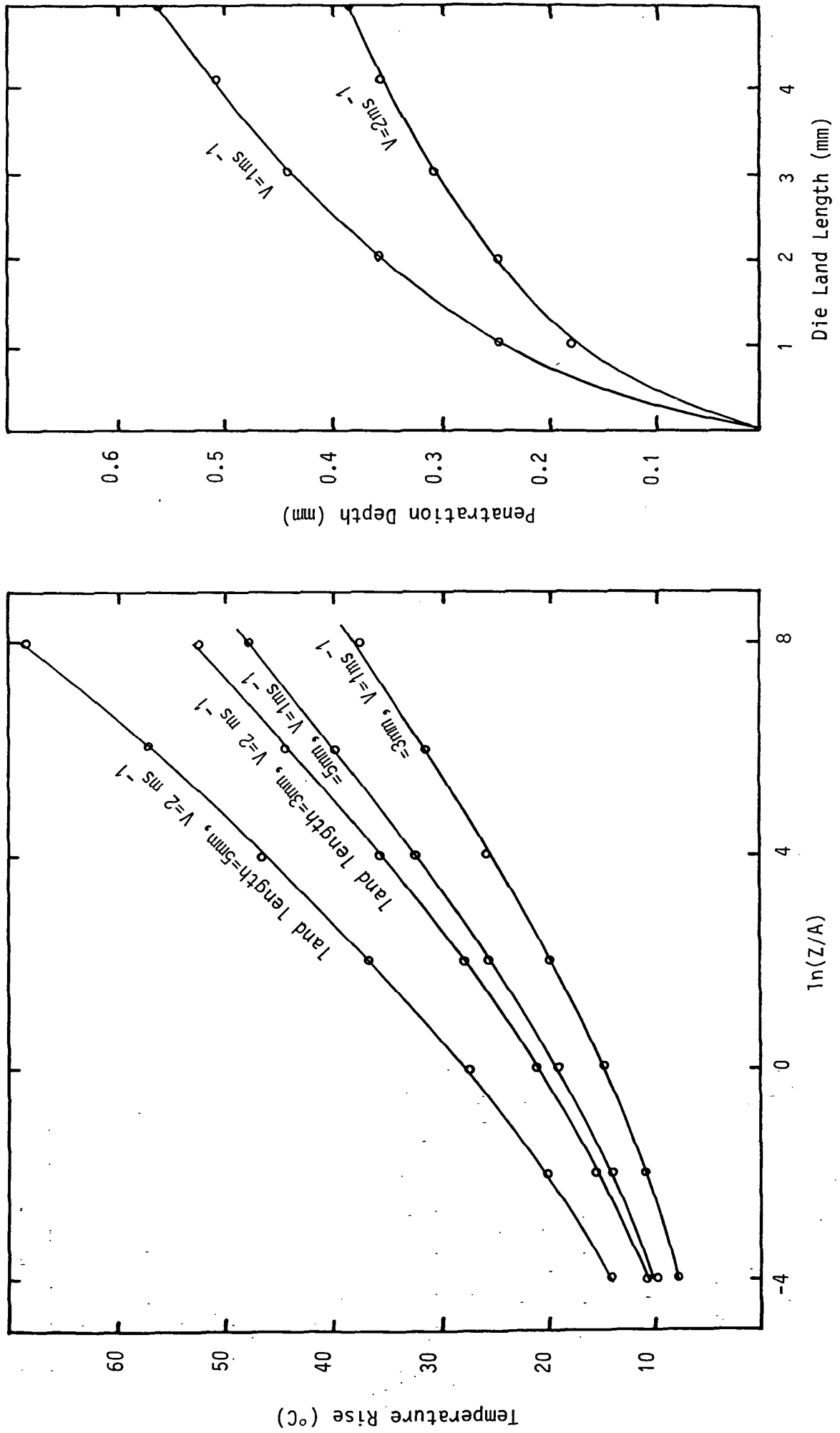


Figure 7.2. Surface Temperature Rise from Die Land Friction

initiate at this location. This is consistent with the assumption made by Lange⁽⁶⁵⁾, (based on experience) that the temperature rise at the edges of a section were approximately twice that at the mid point of the sides.

Examples of a rib section extrudate, from the alloy AA6063, can be seen in Plate (7.3.a.). The extrudate shown at the top of the frame was extruded at 550°C at an exit speed of 1ms^{-1} . The temperature rise from billet deformation and die land interaction was 28°C which is permissible for the alloy in this condition. The second extrudate, shown in the lower portion of Plate (7.3.a.), was extruded at 600°C, and at an exit speed of 1ms^{-1} , resulting in a total temperature rise of 24°C, which was not seen to cause hot shortness for a similar trial with a circular section. Closer examination revealed that the cracking was primarily confined to the corner locations of the section (Plate(7.3.c)), and consisted of a continuous network of incipiently melted metal at the grain boundaries (Plate (7.3.d.)). This confirms the hypothesis that the susceptibility to hot shortness is greater at the corner section of an extrudate. It is also evident that the cracked extrudate has suffered severe pick-up and surface scoring and it will be shown later that it is the onset of this defect that limits extrusion speed and temperature for architectural alloys in the AA6XXX series.

It has been recognised⁽¹¹⁹⁻¹²¹⁾ that the process parameters causing the onset of hot shortness can be predicted using an equation of the form:-

$$\ln Z < a T^{-b} \quad \quad \quad -(7.1)$$

where 'a' and 'b' are constants and 'T' is the initial billet temperature. This equation was applied to the extrusion data for rod section, AA6063, and is shown in Figure (7.3.).

Although Equation (7.1.) is useful in determining the initial extrusion condition resulting in cracking, the exact conditions cannot be properly defined due to the uncertainty of the temperature and stress condition at corner locations of complex sections. Unfortunately there was insufficient data to determine the effect of section shape, on the hot shortness transition line. However, it is clear that the initial billet temperature permissible is lower in complex section.

7.3.2. Tearing

The term surface tearing is often used⁽³⁴⁾ to describe the mechanism by which hot shortness occurs. In this report the term is defined as the partial removal of metal from its matrix to protrude above the extrudate surface.

When an unstable adhesion exists between the die land and extrudate, the tearing defect may occur. The severity of the defect is greatest when extruding through new dies. When new dies are commissioned, the adhesion of the extrudate surface is intermittent both across and along the die land area. This leads to areas of the extrudate material being torn from the adjacent flowing surface. Plate (7.4.a.) shows an example of such severe tearing. It can be seen that the size and geometry of the tear is consistent with the defect

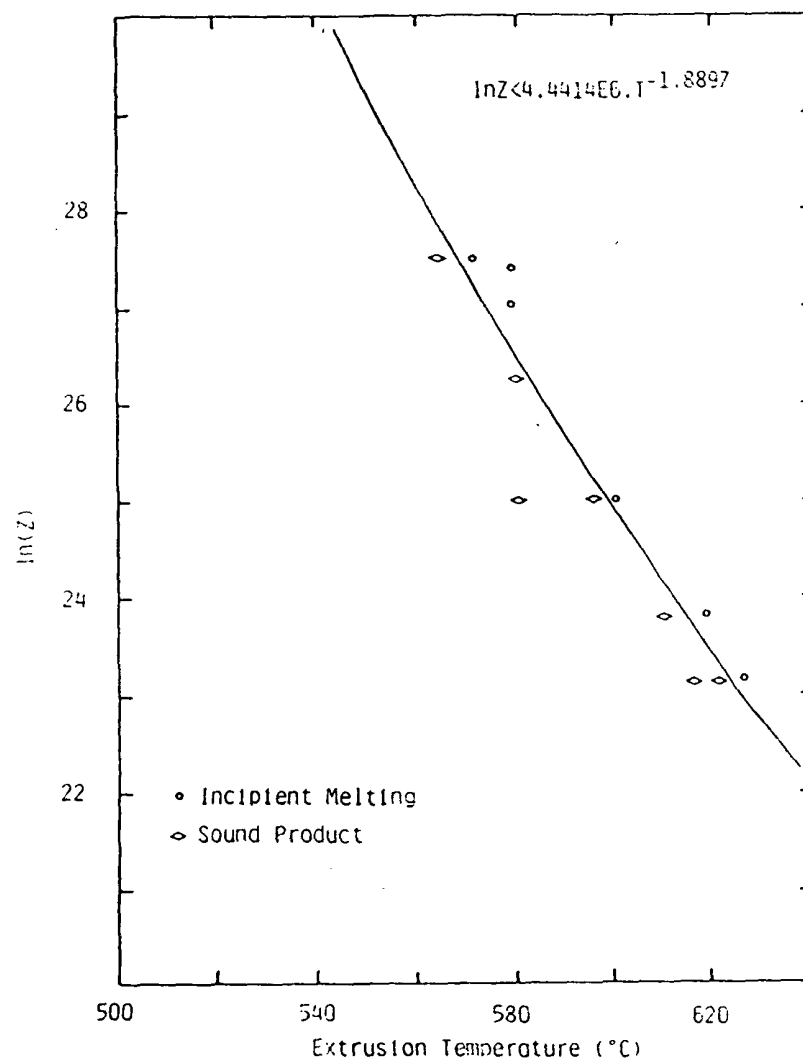


Figure 7.3. Onset of Hot Shortness Vs $\ln Z$

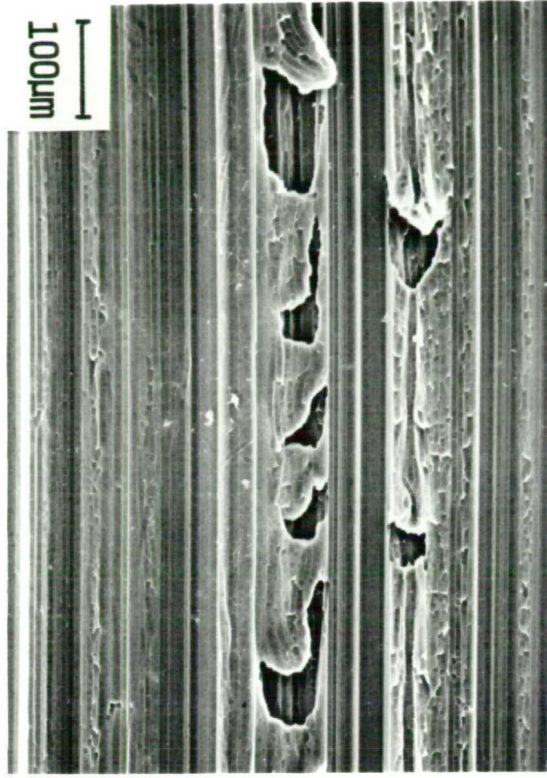
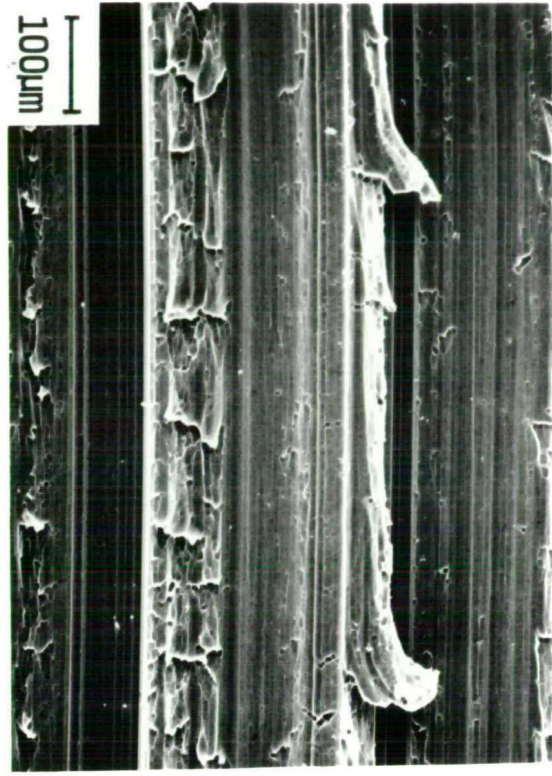
(a) Surface Tearing

(c) Extrudate surface (BEI)

(b) Surface Tearing

(d) Extrudate Surface (BEI)

Plate 7.4. Surface Tearing and Surface Features viewed in SEM.



being associated with a dynamically recovered grain being prised from its boundary by the frictional drag imposed by the die lands. The appearance of the micrograph therefore suggests that the structural integrity of these grain boundaries experiencing excessive shear, on traversing the dead metal zone, is affected by the deformation. It would therefore appear that this type of tearing is associated with both the interaction between the die and extrudate surface, and the strength of grain boundaries.

Because the adhesion between the die land and the extrudate is slightly less than the cohesive strength of the aluminium matrix, the torn grain does not completely dissociate from the extrudate. (It will be seen later that it is this feature that distinguishes it from the pick-up defect.)

Plate (7.4.b.) shows another example of tearing as a result of commissioning a new die. In this event the action of the exit side of the die land has ironed the tears back into the extrudate. Die lines can be seen clearly beneath the ironed protrusions.

The severity of tearing was observed to decrease after the initial extrusions, as a stable layer of aluminium was deposited on the die land area. The coated surface thus eliminates tearing but can be one of the defects known as 'pick-up' and 'die lines'.

7.3.3. Pick-Up and Die Lines

An inherent feature of the extrusion process is that the nature of the

shear involved produces a continuous 'virgin' surface which, in the case of aluminium extrusion, is in intimate contact with the die steel. Moreover, this clean surface is absolutely critical from a commercial consideration. The extrudate/die land interaction involves the development of an aluminium coating on the die land area during the extrusion process, thus producing score lines on the product known as pick-up or die lines. However, in this work die lines were also observed which were not associated with the die land coating. In such cases, the lines were finer and more shallow and termed micro-die lines. In order that the formation mechanisms of the three surface defects might be studied; pick-up, die lines and micro-die lines, specially designed split-die tooling was used. The effect of die land length on the surface formation was also studied using a series of dies with varying lengths from 8mm to 0mm. In the following sections we shall discuss each of the surface defects.

7.3.3.1 Pick-Up

One of the principal surface defects generated during the direct hot extrusion of aluminium and its alloys is 'pick-up'. This defect is observed as intermittent score lines of varying lengths between 3mm and 12mm and often terminate in a fleck of aluminium debris which is referred to as 'pick-up deposit'. As indicated in section 2.8.2., the severity of pick-up is governed by numerous factors, but these may

be grouped into the three main categories:

- (a) Alloy, Condition and Heat Treatment
- (b) Tooling and Equipment
- (c) Extrusion Process Parameters.

In the present study the alloy and heat treatment were kept constant at conditions typical of industrial practice and the object was to determine the effect of extrusion processing. It was not possible to vary the die geometry within the confines of the equipment available except to investigate the effect of die land length on defect production.

The alloy under investigation, AA6063, was supplied in the form of semi-continuous, direct chill (DC) cast logs based on commercial purity aluminium. The alloy's structural cleanliness was similar to that produced for normal industrial processing.

The alloy received a short homogenisation treatment prior to the extrusion trials, which affected iron phase transformation from acicular β to spherical α and produced a fine distribution of Mg_2Si in the β' form. The iron phase transformation being aided by a small amount of Mn in the alloy⁽¹²²⁾. Because of the short industrial type homogenisation treatment, complete spheroidisation of the iron phase was not always achieved and some of the coarser Mg_2Si remained. This enhanced pick-up formation when extrusion conditions were favourable to aluminium build-up on the die land.

The effects of tooling and equipment was not a prime objective in the

present study and so the hot working steel, KEA145, was selected for use throughout the programme. However, the use of new 'non-wetting' die materials such as Syalon⁽⁷⁰⁾ and Zr_2O_3 ⁽⁶⁹⁾ are reported to significantly reduce pick-up by eliminating the aluminium build-up on the die land. The use of such die materials is not yet widespread in the aluminium extrusion industry because hot working steels are, at present, more economic and reliable.

Keeping the factors governing pick-up in categories (a) and (b) constant, the effects of the extrusion temperature and strain-rate could be established. These process parameters were observed to have significant effects on aluminium build-up in the die land area and, as a consequence, pick-up formation.

Before the start of extrusion, the surface layers of both extrusion billet and die land consist of reaction products comprising mostly of oxides. Due to the shear involved in the extrusion process both at the container and dead metal zone interfaces, these protective layers are subjected to shear and indentation resulting in direct metal to metal contact and therefore adhesion at the die location. The extrudate material is thus locally deformed and fragments may be sheared and removed from the matrix and subsequently deposited onto the harder die land surface. Concurrently these deposits are further elongated in the extrusion direction. This appears to be continuous throughout the extrusion cycle producing large areas of transferred material. Eventually, the end zones of the ironed layers on the rear portion of die land begin to spall and a 'pick-up deposit' detaches and becomes embedded in the extrusion product.

Plates (7.5.a,b) show the die land area of a die design for the extrusion of a 25 x 5mm rectangular cross-section bar. The long edge of the die land area is depicted and has experienced three extrusions at 500°C and 575°C in (a) and (b) respectively. The choked area at entry to the die can be seen to have collected a uniform adhered layer of aluminium which was observed to remain stable throughout future extrusion runs. This area is thus responsible for the development of extrudate surface and determines the incidence of the die line phenomenon discussed below and plays little part in the development of pick-up. Following this choked area, the aluminium surface is exposed to the atmosphere and consequently oxidises. It is in the wake of the choked area that flakes of aluminium can be observed to be deposited in the mid section of the die land. Most of these flakes are carried forward by the emerging extrudate and accumulate at the die land exit. Considerable pick-up can be observed on the die at exit; the 'flaked' area of the die being considerably less for the 575°C extrusion which is clearly subjected to greater exit pick-up. The 'pick-up' area in both cases is shown in greater detail by the scanning electron micrographs in Plate (7.5.c,d). Plate (7.5.c) indicates that for the case of the 500°C extrusion the exit area consists of a background deposit of aluminium which probably remains stable but some of the flakes have agglomerated and been ironed as they progress through the die producing a large pick-up deposit, smeared at the tail end which will dissociate from the die and become embedded in the extrusion product. In this case the process is a series of isolated events in which separate agglomerates are successively transferred to the extrudate. In the 575°C extrusion Plate (7.5.d) far more ironing of the particles has occurred and agglomerates have been extended quite substantially in the

(a) Die land area (500°C)

(c) Aluminium build-up on die land
(500°C) viewed in SEM

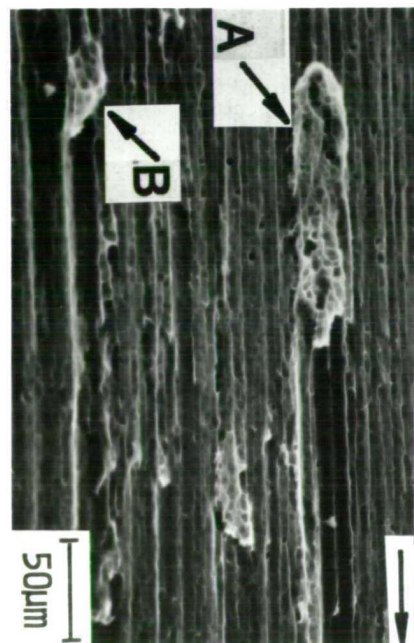
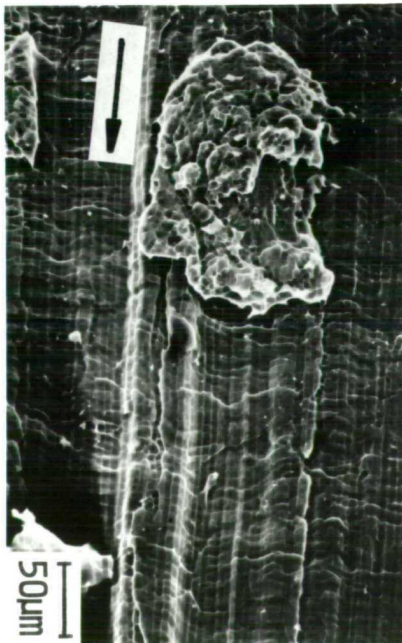
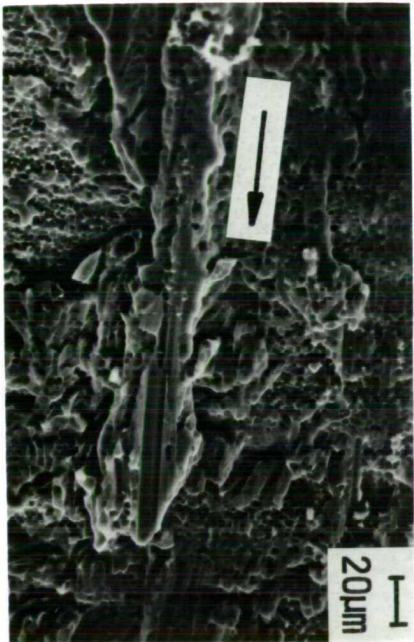
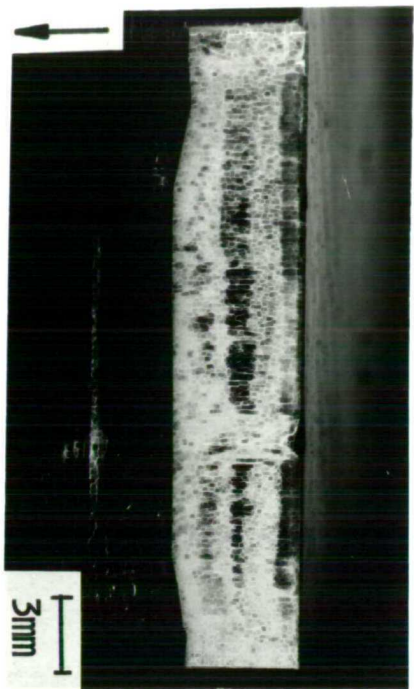
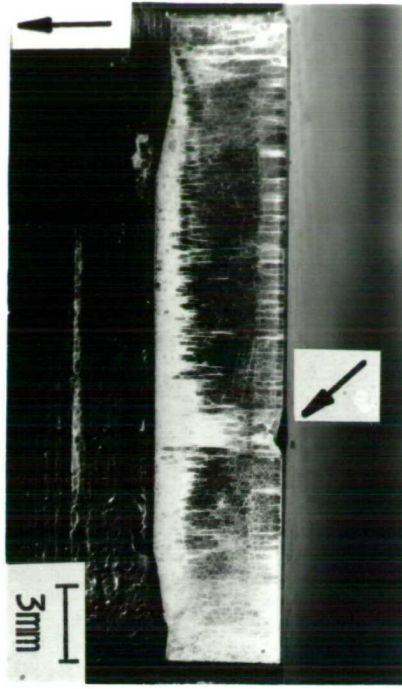
(e) Pick-up (stick-slip mechanism)

(b) Die land area (575°C)

(d) Aluminium build-up on die land
(575°C) viewed in SEM

(f) Pick-up (Debris removal)

Plate 7.5. Aluminium build-up on die lands and the pick-up defect.



extrusion direction; largely because of the lower flow stress of the material. Spalling has occurred either side of the elongated agglomerate which when deposited on the exit side of the die land hardens because of the lower die temperature. Pick-up then occurs by dispersion of the material either side of the elongated agglomerate which will continue to score the extrudate until it is also transferred to the extrusion product. The pick-up damage is thus much greater at 575°C than at 500°C.

In Plate (7.5.a) there is evidence (arrowed) that the throat area of the die has suffered severe damage and as a result a considerable increase in the aluminium film in the wake of the damage. Such damage would lead to immediate withdrawal of the die from service.

We can thus accept that the formation of the pick-up defect is quite clearly due to extrudate/die interaction. The precise mechanism by which this occurs is, however, not so obvious. At least two possibilities exist. The pick-up deposit could simply consist of debris from the coating at exit to the die land or alternatively a mechanism involving transient sticking between extrudate and the die may be responsible for accumulation of extruding material and this could cause the pick-up deposit. The latter mechanism would be analogous to the formation of transfer elements in the wear process. EDX analysis of the pick-up deposit could not differentiate between these two possibilities when single alloy only extrusions were performed. Accordingly an AA5456 billet was extruded at 500°C through a clean die prior to a series of AA6063 trials thus producing an initial deposit of different composition to that of the extrusion billet. The AA6063 billet was then extruded at 600°C at an exit

speed of 0.9 ms^{-1} resulting in both pick-up and surface tearing. The major pick-up deposit indicated as A in the SEM micrograph Plate (7.5.f) was analysed by EDX and found to contain higher magnesium and manganese peaks than would be expected from the parent material, indicating that their origin was from the AA5456 material. However, pick-up deposits with the same element intensities as the parent material (marked B in Plate (7.5.f)) were also observed suggesting that the transient stick-slip mechanism was also operating. After the processing of six billets there was no further trace of pick-up originating from the AA5456 material; a large proportion of deposits having a high percentage of AlFeSi phase.

The results therefore indicate that dual mechanisms are operating. When the propensity for extrudate/die interaction is strong there is a rapid build-up of aluminium at the exit of the die land resulting from severe frictional forces on the aluminium coating causing agglomeration, work hardening, elongation in the extrusion direction and finally spalling which is the ultimate cause of the pick-up on the extrudate. When adhesion between the die and extrudate is approximately equal to the cohesive strength of the metal forming the extrudate surface then the pick-up flakes are continuously transferred between die and extrudate in the die land area, eventually agglomerating as a pick-up deposit. Plate (7.5.e) is most commonly observed when commissioning new dies, where the pick-up deposit consists of metal originating from both the extrudate and the die land coating. An extrusion involving super-purity aluminium produced an excellent example of this mechanism (Plate (7.5)). EDX analysis indicated that the debris, clearly formed by a stick-slip mechanism, to be identical to the parent material.

The stick-slip mechanism is less common and results in less severe extrusion damage than occurs when the deposit is produced by continuous accumulation at the die exit. In the latter case the damage is generally observed as a deep score line with the pick-up deposited at its termination as shown in Plate (7.5.f) labelled A.

The occurrence of excessive pick-up is a matter of prime concern especially where cosmetic features are commercially important. It is therefore necessary to attempt to quantify the phenomenon, perhaps in a similar manner to that derived for acceptable surface in the high strength aluminium alloys⁽¹²³⁾. The severity of pick-up increased with temperature and strain-rate which coincided with a greater degree of aluminium build-up on the die land. The process conditions at which excessive pick-up occurred (i.e. gross product damage) should thus be related to the temperature-compensated strain-rate which is defined as $Z = \dot{\epsilon} \exp (\Delta H/RT)$ where $\dot{\epsilon}$ is the strain-rate, T the extrusion temperature, ΔH an activation energy and R is the universal gas constant. Moreover, the flow stress of the material is a function of $\ln Z$. Hence Figure (7.4.) shows the function $\ln Z$ plotted against the initial billet temperature T . The locus dividing the experimental results exhibiting an acceptable surface and those showing excessive pick-up, are clearly divided by the inequality $\ln Z < a T^{-b}$.

It is likely that the values of 'a' and 'b' in this equation are unique to the conditions utilised in these experiments, for example, alloy, cast condition, homogenisation treatment, tooling design and tooling temperature. However, it is likely that the functional form

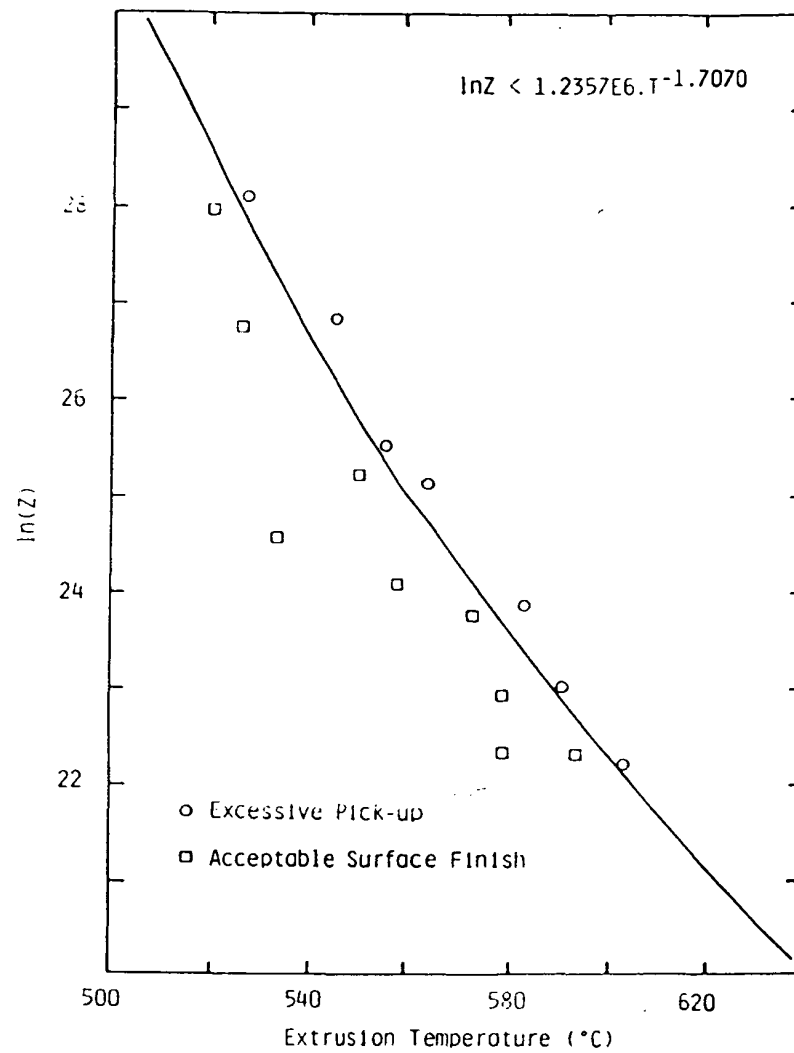


Figure 7.4. Onset of Excessive Pick-Up Vs $\ln Z$

would be applicable to most extrusion conditions and provides a useful relationship which, when used in conjunction with a microcomputer, could rapidly be utilised as an industrial tool.

7.3.3.2. Die Lines

The Aluminium Association has defined die lines as "longitudinal depressions or protrusions formed on the surface of the drawn or extruded material due to imperfections on the die surface". Although extrudate/die interaction is clearly a major cause of the die line formation, an alternative mechanism of generation was observed in the present work giving rise to micro-die lines. The two types of die line generation operate simultaneously during extrusion resulting in a surface with combined features. Therefore the complete surface topology determining brightness, reaction to anodisation etc. consists of at least two components.

Before differentiating between the two types of die lines it is useful to study the global features of the extrudate surface finish. In order to achieve this, trials were performed at temperatures of between 350°C and 550°C and at an exit speed of 1 ms⁻¹. Three billets were processed at each temperature. The die lands were cleaned prior to the first extrusion at each temperature and the surface finish measured using a stylus type instrument. Measurements were made directly in the case of the split die tooling and using replica resins when conventional dies were used. The surface finish in each case was approximately 0.5µm Ra.

The extrudate surface roughness was sampled at 0.5m intervals along

the product using the replica resin. Because of the variation in the Ra values at each test temperature, they are presented as scatter bands in Figure (7.5.). The plot indicates that there is an optimum operating temperature range from 425°C to 475°C in which the smoothest surface finish is obtained. Also shown in Figure (7.5.) is the micro-die line roughness which is discussed in the next section.

The variation in the surface roughness can be directly related to the morphology of the aluminium film on the die land area. This film has two characteristic features. At the front portion of the die land the newly formed extrudate chokes the die throat. The extrudate surface and the die land are in intimate contact preventing exposure to the atmosphere. After leaving the choked area of the die land, the small amount of die deflection causes the extrudate to dissociate from intimate contact permitting oxide formation on the surface. The interaction between extrudate surface and die land then creates the flaking and aluminium film development to which pick-up can be attributed. Therefore, the final surface formation mechanism may be regarded as one of generation and modification. The surface is generated at the forward position of the die throat in the absence of oxygen. Once generated, the rear portion of the die land serves only to deteriorate the surface by the imposed ironing and galling of the extrusion product. It is the relative contribution of the two areas on the die land that governs the surface finish at each temperature condition.

At the high temperature condition (550°C) the choked aluminium at the entrance to the die throat adheres and produces a band like formation flowing into the valleys of the ground surface. (See Plate (7.6.)).

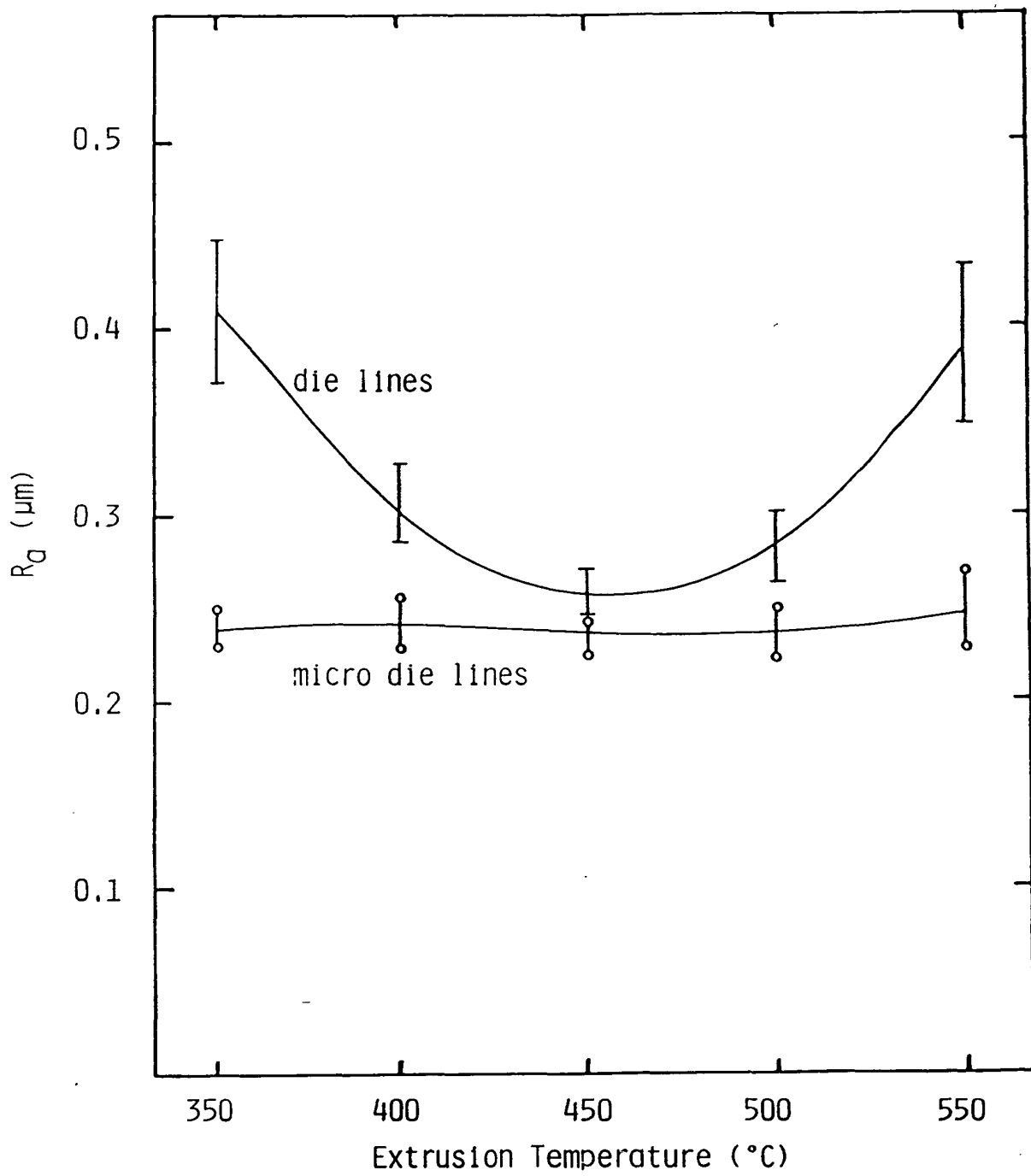


Figure 7.5. Die Line Roughness Vs Extrusion Temperature

The banded aluminium film becomes harder than the adjacent extrudate material, due to die quenching and work hardening. Because of the low flow stress of the material the die valleys overflow and ridges of hard aluminium are present in the choked area. The surface layers of the extrudate are therefore formed by the ridges of the aluminium film at this location. As the extrudate leaves the choked position the surface possesses the topographical features of the aluminium film. The continued interaction with the remainder of the die land creates an unstable aluminium film, accumulating at the exit side of the die, which scores the surface and could also cause pick-up. The aluminium coating at the front of the die land is more stable and does not appear to alter significantly with extrusion runs.

In the mid-temperature range (425°C to 475°C) the aluminium film appears more speckled at the choked portion of the die land. This can be seen in Plate (7.6.b). In this range, the aluminium is in immediate contact with the die land and resides in the ground surface topography but unlike the high temperature condition, does not protrude above the original ground surface. This effectively reduces the surface roughness of the die land. The die is now effectively perfectly smooth and permits sliding of the surface without deleterious high friction conditions thus producing a smoother product. The reduced aluminium film on the leading portion of the die land is probably due to the scouring effect of the harder Mg_2Si particles, now finely distributed throughout the matrix.

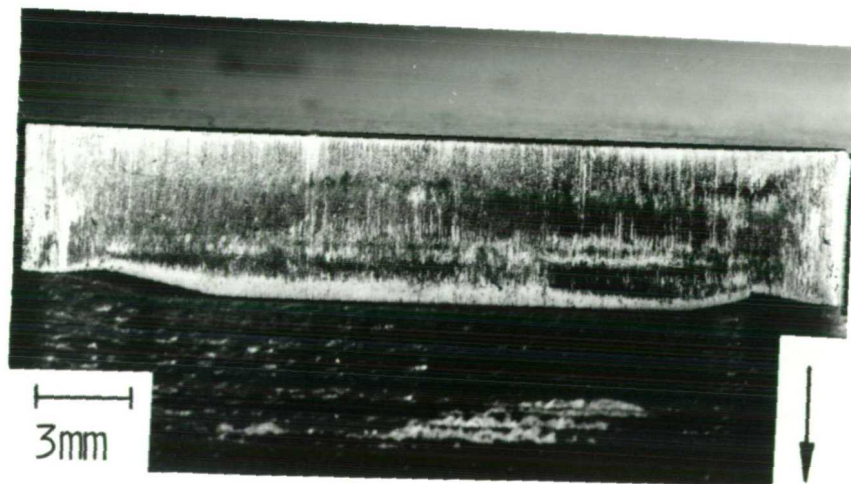
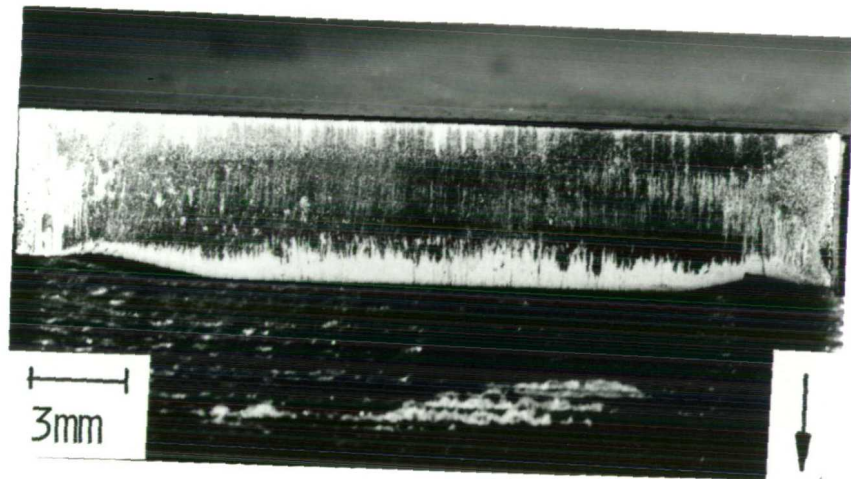
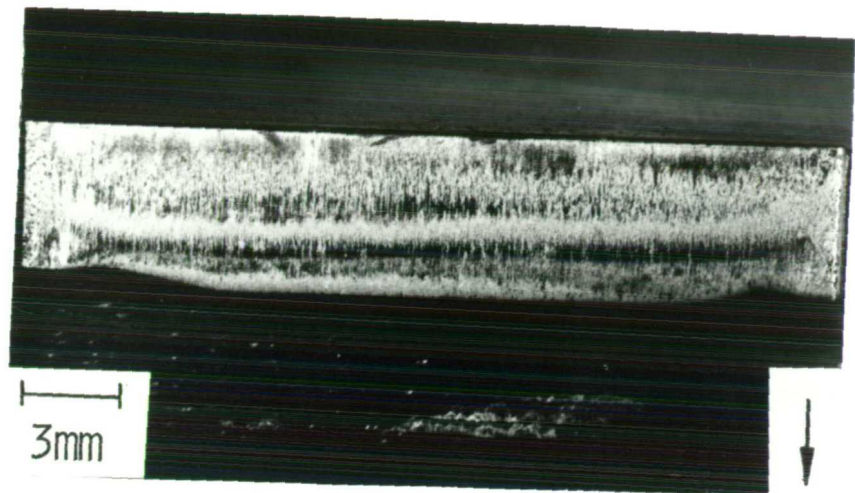
There was also a build-up of aluminium on the exit side of the die land which must have increased the extrudate's surface roughness. However, the film in this case was more stable than the high

(a) 550°C

(b) 450°C

(c) 350°C

Plate 7.6. Aluminium Build-up on Die-Lands.



temperature condition and did not appear to cause pick-up.

At the lowest temperature condition studied (350°C) an increase in the surface roughness was observed and was accompanied with a greater scatter of results. At this temperature the rejection of Mg_2Si is high, forming coarse precipitates and hence largely eliminating any scouring effect. Indeed the possibility of such acicular precipitates lodging in the valleys of the die surface cannot be rejected. The reduced temperature also caused an increase in the flow stress of the metal producing higher extrusion pressures and increased die deflection, which is known to enhance build-up of aluminium on the die land⁽⁷¹⁾. This feature is clearly illustrated in Plate (7.6.c). The aluminium coating in the choked area has increased slightly and coarse bands of aluminium have accumulated, aligned in the extrusion direction. This film was formed in the absence of oxygen and unlike the higher extrusion temperatures, there was less resultant build-up of aluminium on the exit side of the die. Therefore, the product surface was produced by further extrusion through the choked aluminium film, thus adopting its coarser topographical features.

It is evident from the results presented in Figure (7.5.) and the observations made on the aluminium coating on the die land areas, that the die land/extrudate interaction plays an important role in determining the surface finish of the extrudate. It may also be observed that the surface is generated in the choked area and the remaining die land deleteriously modifies the product finish. The effect of die land length on product quality is thus quite clearly an important aspect.

From the trials discussed above there is an apparent optimum operating temperature in the vicinity of 450°C, which is consistent with industrial practices for AA6063 extrusion. To investigate the effect of die land length, the extrusion parameters were kept constant at 450°C and an exit speed of 1 ms⁻¹, and a series of trials were performed using identical profile dies (20:1 strip), with land lengths varying from 8mm to knife edge (0mm). The dies were ground perpendicular to the extrusion direction with a surface finish of 1.1µm Ra. Three billets were processed through each die in turn and the results are shown in Figure (7.6.).

The initial billets processed through the dies exhibited a rough topography and displayed a high percentage of tearing; the roughness and tearing increasing with the die land length. This feature is commonly observed when commissioning new dies because the adhesion between the newly generated extrudate surface and the clean, ground die land is intermittent both across and along the die land area. This leads to areas of the extrudate material being torn from the adjacent flowing surface, then ironed into the extrudate surface, as it continues to traverse the die land. An illustration of the severe tearing observed is shown in Plate (7.4.).

The third and final extrusion run of the series produced a slightly smoother surface than the second process; the effects of the die land length being similar. The coating of aluminium on the die land in this case was more fully established; residing in the grooves of the die land surface.

An interesting feature observed after the trials was that the build-up

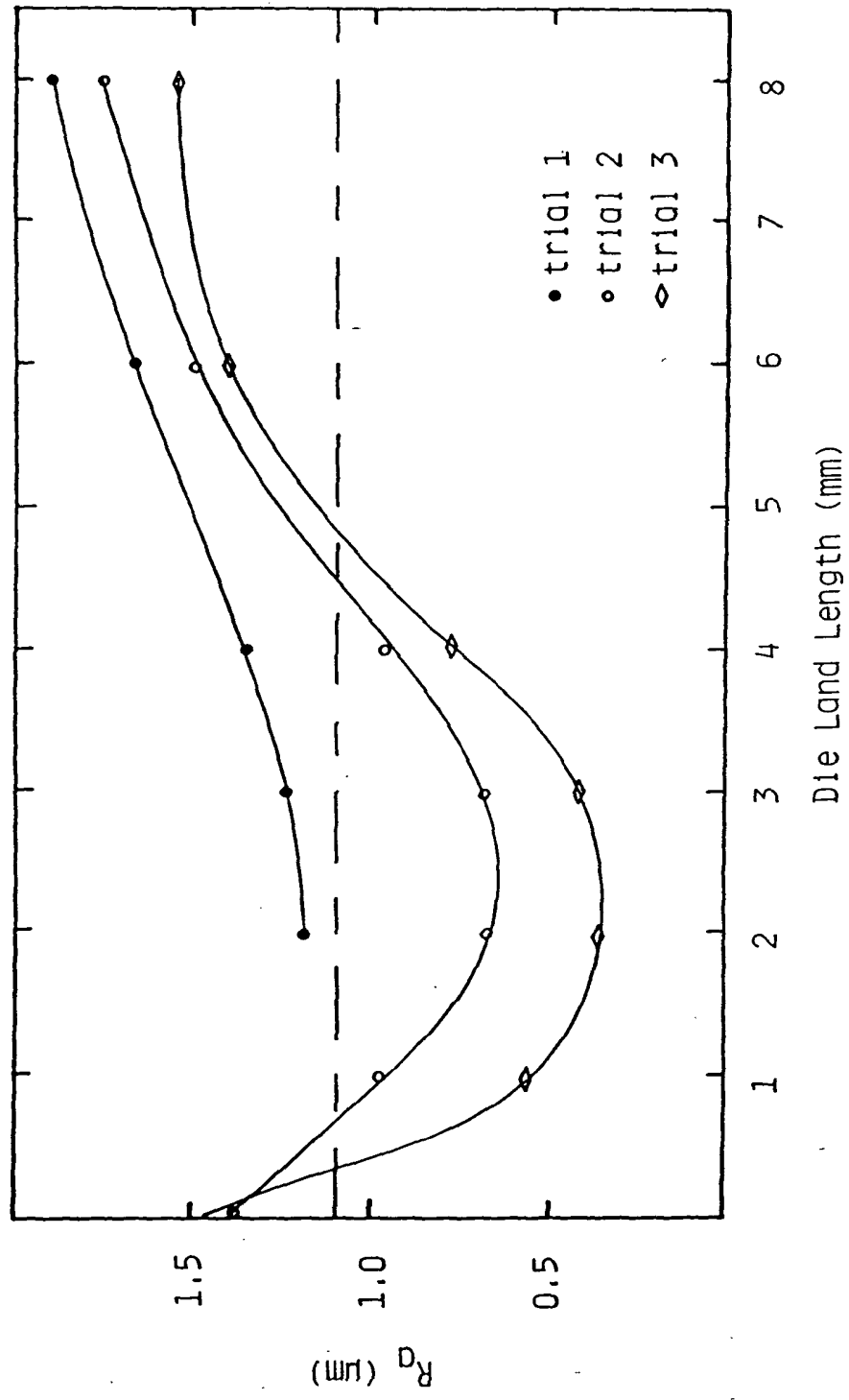


Figure 7.6. Die Line Roughness Vs Die Land Length

on the working area of the die land (i.e. the front portion) was about 2mm in length for each of the varying die land lengths. This provides further evidence that the surface is generated at the entrance to the die throat and that the remainder of the die land towards the exit side serves only to increase friction and provide greater surface area for material transfer and hence a rougher product and increased likelihood of pick-up.

From these observations it might be expected that the extrusion trials performed using the knife edge die would exhibit an excellent finish with no pick-up or die lines. However, it was shown in Figure (7.6.) that the surfaces produced were rougher than those of 2mm and 3mm die land lengths.

The reason for the deterioration in the surface quality is primarily due to the instability of the knife edge under the extrusion load, causing a chattered surface appearance. There was also a considerable amount of build-up in the relief behind the knife edge which was associated with surface scoring.

Therefore it becomes evident that there is a minimum die land length (in this case about 2mm) for successful surface generation. At this optimum condition the entire land length is choked preventing the introduction of oxygen to the virgin surface; the prevention of an aluminium oxide film which has the propensity to score the extrudate surface is most clearly a critical feature.

7.3.3.3. Micro-Die Lines

The discussion above has indicated that to achieve the best possible surface finish, extrusion must be performed at about 450°C with the die in the choked condition. However, even at these optimum conditions, die lines were still observed to occur. Polishing the die lands had some effect; reducing their occurrence and improving the surface quality. However, it did not eliminate this defect.

It was shown earlier in Figure (7.5.) that the micro-die line roughness, measured by interferometry, was almost constant at $R_a = 0.225\mu\text{m}$. This suggests that micro-die line generation is athermal. Closer examination of the lines revealed that they were both finer and less deep than die lines produced by the mechanisms discussed above. Clearly their origins lay in some other mechanism hitherto not investigated.

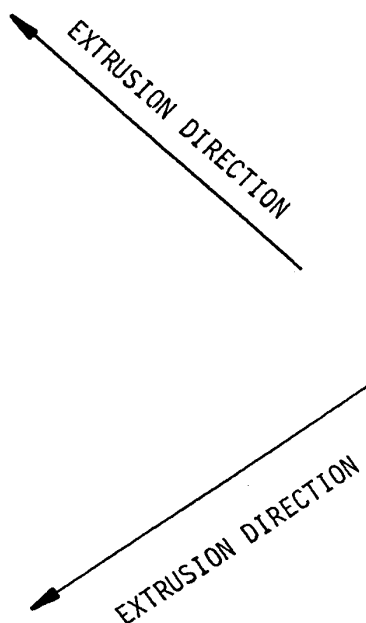
When the extrudate surface produced under these conditions was observed under the SEM, in the BEI mode, elongated iron phase precipitates could be identified in the subcutaneous layers and were clearly associated with surface scoring. Plate (7.4.c,d) are examples of the as-extruded surface viewed by back scattered imaging. The iron phase which appears white in the micrographs can be identified in the sub-surface layers, aligned in the extrusion direction. It is these precipitates that appear to be associated with the fine score lines on the extrudate surface, which will henceforth be termed micro-die lines. One other interesting feature of the back scattered imaging utilised in Plate (7.4.c,d) is that the recrystallised grain structure may be discerned.

Because of the low resolution of the SEM, in the BEI mode, at magnifications greater than 1000X, an alternative method of studying the features of such micro-die lines was clearly required. A technique of preparing specimens from the extrudate surface for TEM examination was developed which involved removing a slice 300 μ m thick from the extrudate surface using a diamond wafering wheel. 3mm discs were then punched from the slice and subsequently electropolished in a jet thinner from the bulk extrudate side only. One-sided polishing was achieved by placing an acetate washer between the extrudate surface side of the disc and the specimen holder. The acetate washer was necessary in order that the jet thinner's detector would still function.

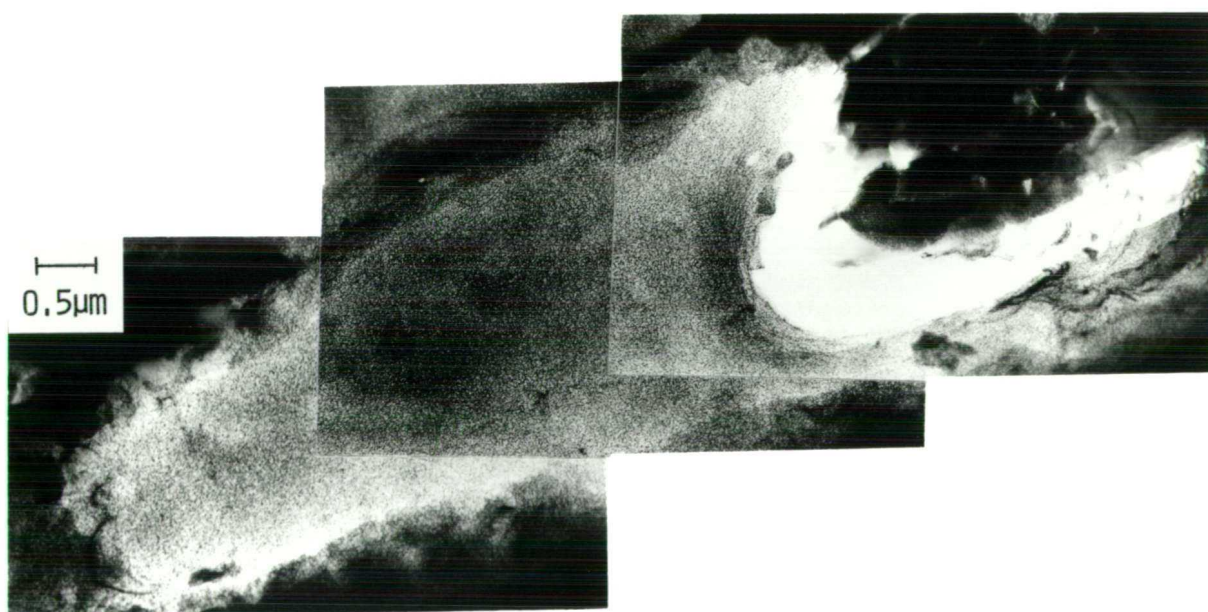
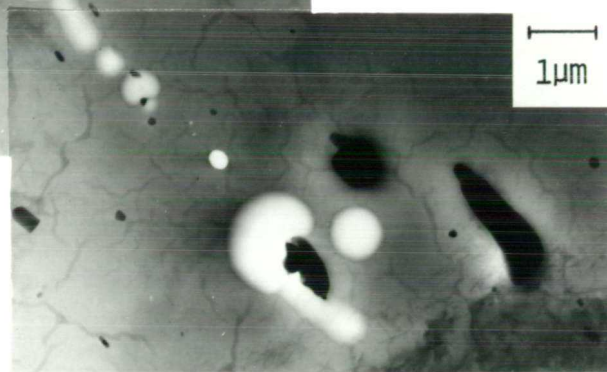
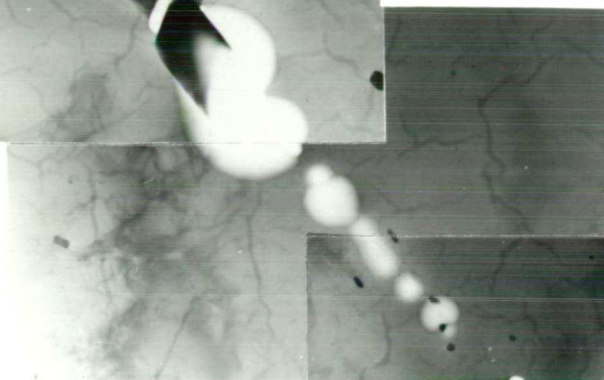
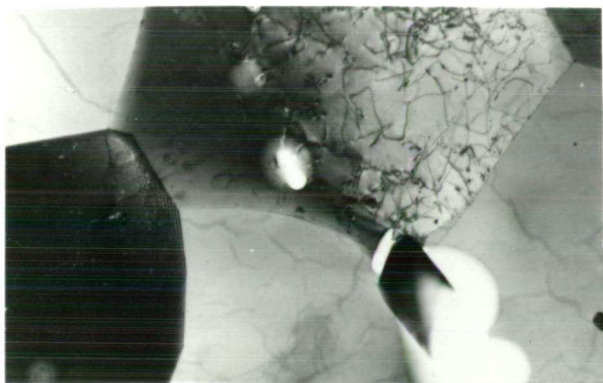
Observation of the electropolished foils in the TEM made it quite clear that the iron phase precipitates identified earlier were in fact fragments of larger precipitates interspersed with considerable cavitation. The cavitation links each fragment linearly and it is these strings of voids which appear as micro-die lines on the extrudate surface. One example of this mechanism is illustrated by Plate (7.7.a) which also clearly shows the 'press quench' effect with sub-grains clearly visible in the top left hand corner of the micrograph.

In addition to the micro-die lines some die lines were also observed originating from the interaction of the extrudate and die land area. Plate (7.7.b) shows one example of a die line termination point. The cause of this surface score was a build-up of fragmented iron phase particles in the aluminium film on the die land. As extrusion

- (a) Fragmented Iron Phase in the Surface Layers
of an AA6063 Extrudate.



- (b) Iron Phase Cluster in the Surface Layers
of an AA6063 Extrudate.



proceeded the iron phase cluster spalled, detached from the die land, and transferred to the extrudate's surface layers.

Plate (7.8.) provides further proof of the role of the fractured iron phase in surface generation. To obtain this evidence it was necessary to prepare the specimens for TEM by electropolishing from both sides of the ground foil. This allowed additional features of the precipitates residing in the subcutaneous layers to be identified. It became clear that, in addition to the fragmented coarser AlFeSi precipitates, bands of small spherical precipitates were also present (Plate (7.8.a,e)). The SADP shown in Plate (7.8.c) identifies these precipitates as AlFe(X)Si whose morphology was not observed elsewhere in the extrudate being quite specific to the subcutaneous layers. These particles, as indicated in Plate (7.8.d), originate from the disintegration of the coarser precipitates and were observed predominantly at entry to the die throat area where the extrudate experiences severe shear deformation. The morphology observed in Plate (7.8.e) suggests that the particles originate from that part of the material which is located near to the flat face of the die, on the edge of the dead metal zone and traversing more slowly into the extrudate surface. Pressure-assisted capillarity-driven diffusion process would quite rapidly transform fragmented precipitates into a spheroidised form at such locations. Moreover such spheroidised fragmented particles could not be identified in any other location in the deformation zone. The small precipitates proceed to the die throat/extrudate interface and are released into the sub-surface layers. The considerable dislocation activity at these precipitates Plate (7.8.e,f) affects the surface topology by rearranging into an equilibrium array and eventually coalescing to form recrystallisation

(a) Coarse AlFeSi and Spherodised particles

(b) Distribution of precipitates in subcutaneous layer

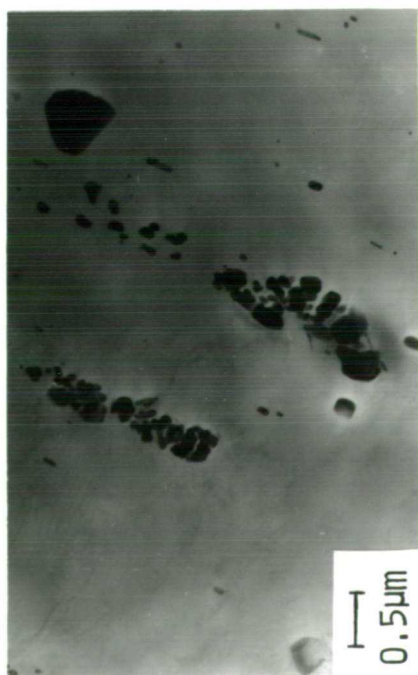
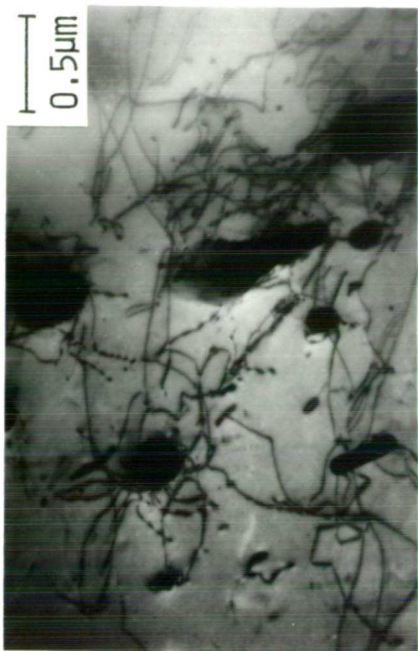
(c) SADP of spherodised particles

(d) Disintegration of AlFeSi

(e) Spherical precipitates in subcutaneous layers

(f) Dislocation activity at precipitates

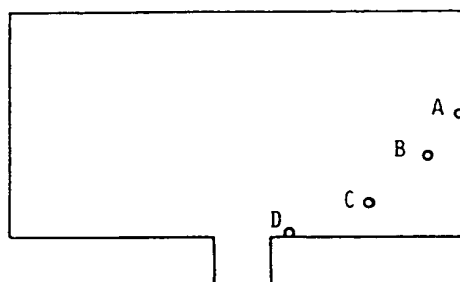
Plate 7.8. Subcutaneous Features of an AA6063 Extrudate.



nuclei. It appears that these small precipitates are almost completely released into the sub-surface layers forming the dispersion shown in Plate (7.8.b) presumably due to the retarded flow of material close to the flat face and into the radiused choke area of the die. This evidence clearly suggested that the internal shear/dead metal zone boundary was effectively the origin of all micro-die lines and hence a thorough investigation was conducted at this location. Plate (7.9.) shows the micrographs observed and their precise location in the deformation zone. At entry to the shear zone (location A) the iron phase precipitates commence to align themselves into the direction of flow concomitantly pinning sub-grain boundaries: the apparently spherical precipitates shown in Plate (7.9.b) are of course the cross section of needle type precipitates observed in cross section. At this position there is little evidence of fragmentation of the precipitates although the volume fraction might suggest that some of the precipitates originate from even larger particles residual from the homogenisation process. As the particles approach the die throat, however, it is clear that they are elongated, sheared and become aligned into the high shear direction (Plate (7.9.c,d)). Plate (7.9.e) shows the fracture of this iron phase with the associated stress field generating dislocations between the individual remnants of the original particles. This is shown at higher magnification in Plate (7.9.f). This stress field between the fragments eventually leads to the cavitation reported above which is the direct cause of the observed micro-die lines.

It was discussed above and shown in Plate (7.8.), that bands of spheroidised iron-rich particles may be detected concomitant with the fragments of broken acicular phase in the subcutaneous layers of the

(a) Quasi-static flow



(b) Location A

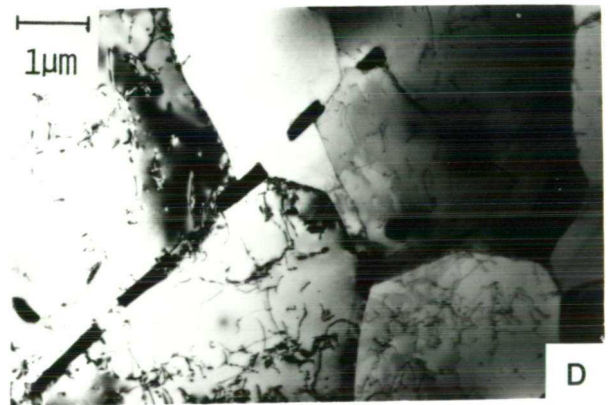
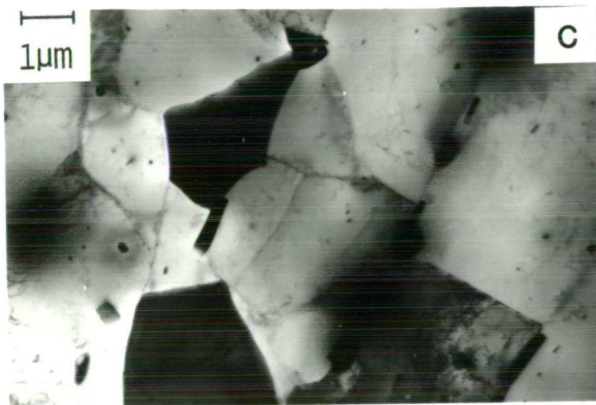
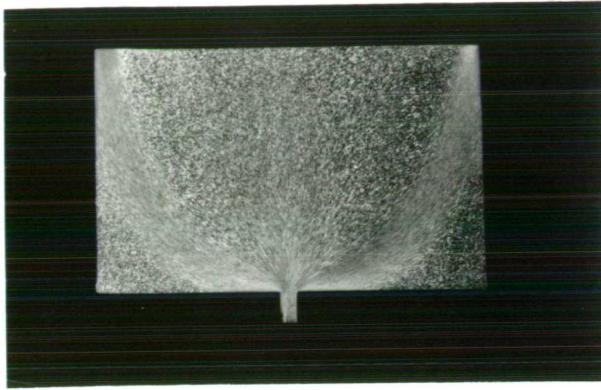
(c) Location B

(d) Location C

(e) Location D

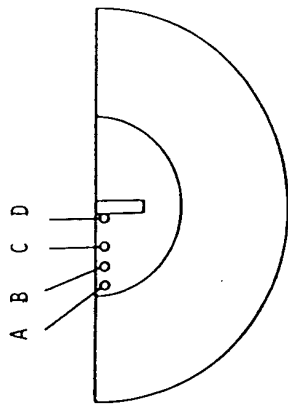
(f) Location D'

Plate 7.9. Structural features in deformation zone
(meridian plane)



extrudate. In an attempt to determine the prime origin of these spheroidal particles the die face adjacent to the die throat was examined in more detail by transmission electron microscopy. Plate (7.10.) shows the results obtained. The macrograph (7.10.a) indicates that the periphery of the face is composed of coarse recrystallised grains and is truly quasi-static as is the region spreading in from the outer edges to about $1/3$ of the billet diameter composed of much smaller recrystallised grains. The core of the material appears to be heavily deformed with material flowing into the die throat. The extrusion was again rectangular and the positions from which specimens were extracted for examination in the TEM are indicated in the diagram (Plate 7.10.). It should be noted that the material along the die face is feeding into the die throat much slower than the material in the shear zone adjacent to the dead metal zone. It is in this volume, at the die face, that coarse fragmented particles congregate, awaiting release into the extrudate surface. Clearly the volume fraction of dispersoid is much higher as indicated in Plate (7.10.b,c) which shows a well formed sub-structure, the boundaries being pinned by small Mg_2Si particles with quite large precipitates being observed elsewhere. It is likely that, at this stage, the precipitates are not aligned into the flow direction and hence the micrograph is showing most of the particles in cross section. It should also be noted that the considerable dislocation activity within sub-grains, suggests that flow is still developing in this region. The precipitates observed are mainly Mg_2Si caused by the billet reaction with the cooler die interface ($350^{\circ}C$). In regions closer to the die throat entry there is an increase of the iron phase fragments which become more aligned as the material moves from location [B] to location [C] (i.e. Plate (7.10.c to 7.10.d)).

(a) Quasi-static Flow
(billet face)



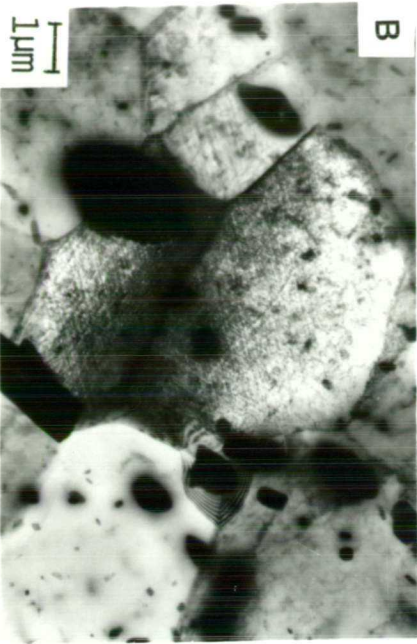
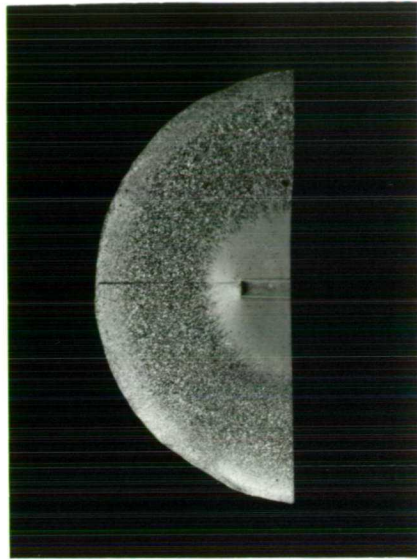
(b) Location A

(c) Location B

(d) Location C

(e) Location D

Plate 7.10." Structural Features at Billet/die Interface.



These fragments are clearly deposited by the intense shear zone moving along the dead metal zone boundary. In addition to the alignment occurring between [B] and [C] some spheroidisation also occurs due to the shear—assisted capillarity processes discussed above and similar to that described by Zaidi and Sheppard⁽¹⁸⁾. The fragmentation of the residual large particles just before entry to the die is clearly illustrated in Plate (7.10.e) from the specimens extracted from location [D].

It is thus clear that the surface of the extrudate originates from two distinct regions of the material. The major contribution is from the deformation zone of intense shear, decorating the dead metal zone. The smaller volume moving more slowly along the die face collects the mainly fragmented iron phase, assists in spheroidisation of this phase and is also responsible for the introduction of small Mg_2Si precipitates into the surface. Thus the origin of the micro-die lines can be attributed to the major region of intense shear but probably would not operate efficiently without the aid of the die face deformation zone. Thus there is clearly scope for revision of die designs in an attempt to solve this problem.

7.4. Extrusion Topology (Limit) Diagrams

Extrusion limit diagrams have been reported^(124,125) which are able to indicate operating conditions to predict strength, toughness, elongation, acceptable surface, etc. When processing architectural alloys such as AA6063, the production requirements and quality of extrudates is dictated mainly by cosmetic considerations. For such

alloys it would be helpful to present the surface generated features within the limit diagram framework.

The limiting criterion for an acceptable/unacceptable surface condition on the original limit diagram⁽¹²⁶⁾ was when incipient melting occurred. Alternatively, the onset of hot shortness, due to insufficient tensile strength at the extrudate surface, can be used to represent this limit. When cosmetic features are the prime requisite of extrusion, it is the incidence of excessive pick-up that defines the extrusion limit.

It was shown in section 7.3.3.1 that the conditions for excessive pick-up can be presented in the form:

$$\ln Z < 1.2357 \cdot 10^6 \cdot T^{-1.7070}$$

Thus it is possible to represent the results on the extrusion limit diagram. Figure (7.7.) shows such a diagram for AA6063. The pick-up line appears in the operating area of the diagram adjacent to the incipient melting line. It is clear that it is this limit that governs the operating area for architectural alloys. The limit conditions for the onset of excessive pick-up are additionally dependent upon the process non-variables such as cast cleanliness, homogenisation treatments and extrusion tooling and practices.

Therefore, the limit line may vary from one extruder to the next and between batches of the cast alloy. However, it is clear that the form of the representation will be consistent.

The most common surface defect in aluminium extrusion is die lines.

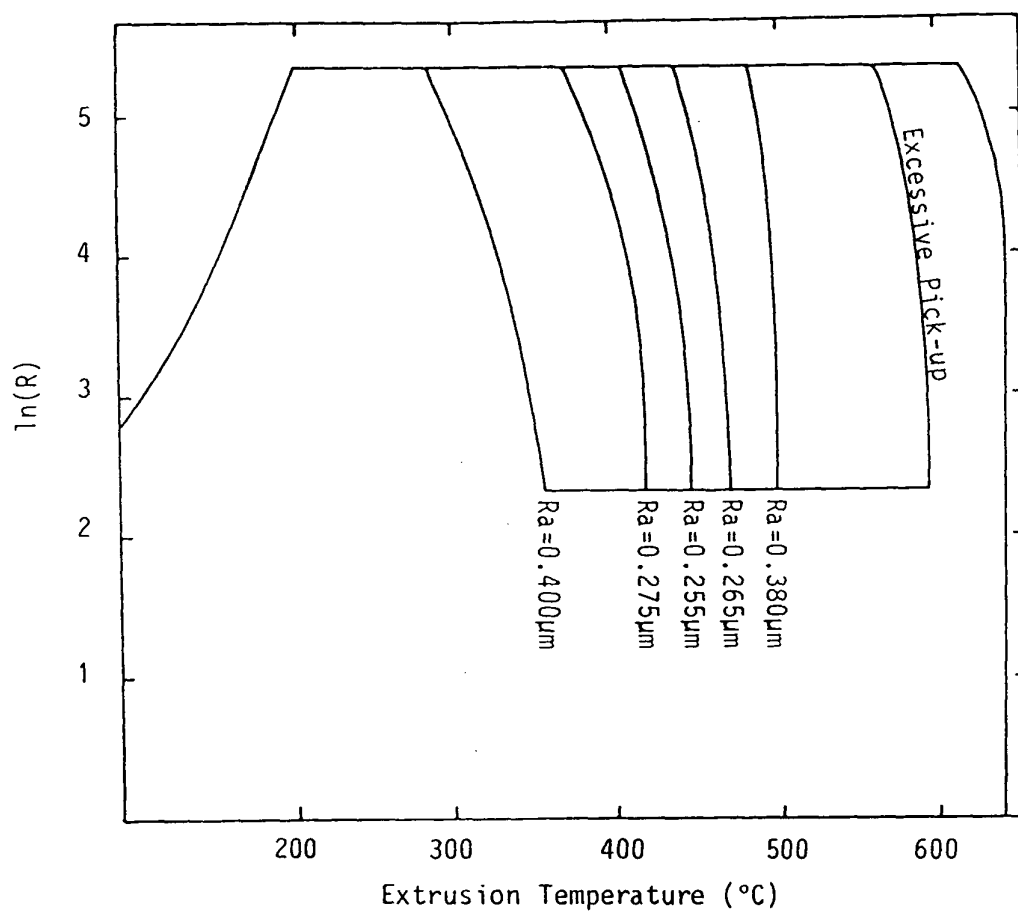


Figure 7.7. Extrusion (Topology) Limit Diagram
for AA6063

which appear at all processing conditions. Although die lines are termed a surface defect, they are tolerated in many high strength alloy extrusions as they do not appear to affect mechanical properties. However, die lines prove to be most deleterious in architectural alloys especially where anodising treatments are to be used. Inclusion of such information within the limit diagram framework should therefore prove most useful. It was shown in section 7.3.3.2. that the average die line value (R_a) varied with temperature and temperature-compensated strain-rate. This allowed conditions for constant surface roughness to be formulated. As these results are presented in a quantitative form it is possible to include them on the extrusion limit diagram. Figure (7.7.) shows an extrusion limit diagram for AA6063 with contour mapping of constant die line roughness. The high degree of scatter in surface roughness values at each process condition indicates that employment of such a representation of results would be of limited practical value. A more useful form of the diagram is one displaying operating areas in which an extrudate surface roughness within a given R_a band might be obtained. This form of diagram is shown in Figure (7.8.) which was constructed from the same results presented in Figure (7.7.).

Clearly, these results were obtained under laboratory conditions where control of die conditions and process parameters were relatively simple. Consequently, it is expected that industrially there will be some variation in results when extruding varying sections. However, the general form of the equation should enable the extrusion press operator to generate sufficient data to correlate with his industrial results. Almost certainly, the information would be held in a micro-computer and the data used directly rather than by construction of diagrams.

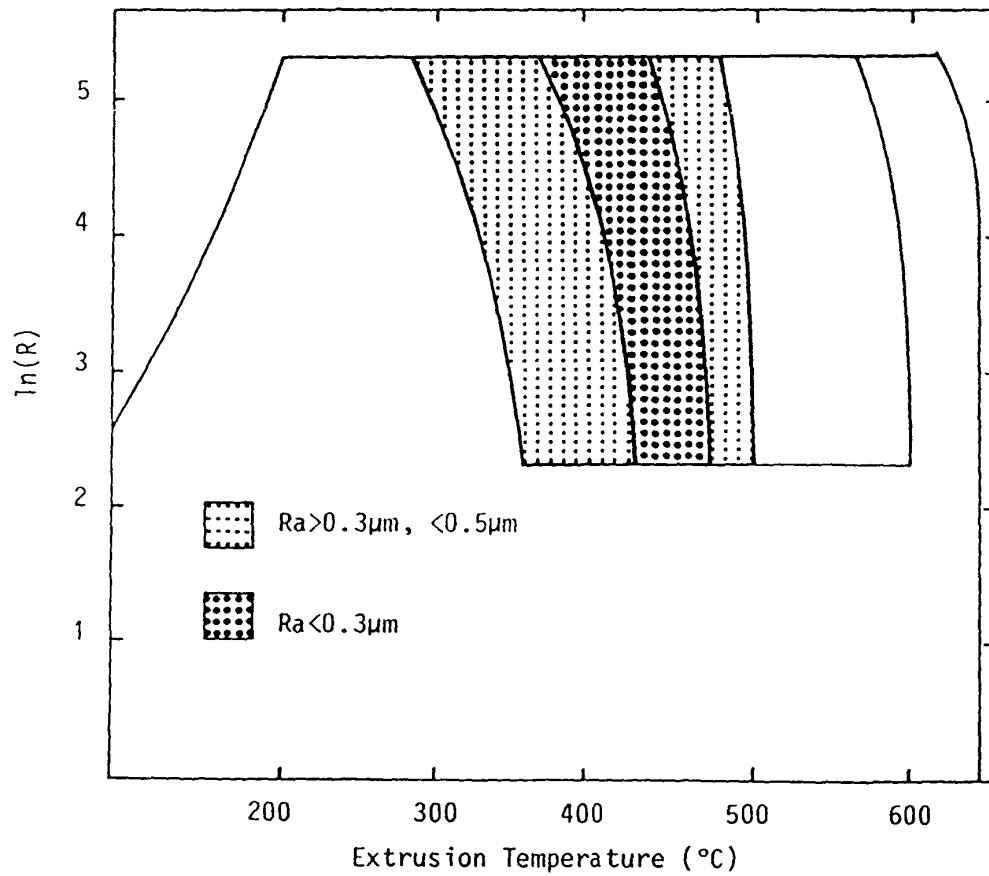


Figure 7.8. Extrusion (Topology) Diagram with Operating Regions

CONCLUSIONS

- (1) In non-lubricated, direct extrusion, through flat faced dies, a virgin surface is generated by shearing from within the extrusion billet
- (2) Tearing results from a state of intermittent adhesion between the extrudate and the die land. This defect is most common when commissioning new dies.
- (3) The occurrence of hot shortness and the onset of excessive pick-up can be controlled using the inequality:

$$\ln Z < a.T^{-b}$$
- (4) The surface formation mechanism in extrusion is one of generation and modification : the surface is generated at the choked portion of the die land and is subsequently modified by interacting with the remainder of the die surface.
- (5) Homogenisation treatment is essential to reduce excessive pick-up by transforming the inter-metallic iron phase from β to α and redistributing the Mg and Si.

- (6) The generation of the pick-up defect occurs by two mechanisms;
- (a) transient sticking between extrudate and die land resulting in accumulation of extrudate material. (Analogous to the formation of transfer elements in the wear process).
 - (b) debris removal from the aluminium coating on the die land area.

The latter mechanism is the most common in AA6063 extrusion.

- (7) Die lines are largely the result of interaction between the extrudate and the choked portion of the die land.
- (8) Optimum process conditions can be established, together with a minimum die land length. The optimum die land length is that which gives sufficient stability together with a land length which is totally choked.
- (9) Operating at optimum conditions, and with highly polished dies, continues to generate die-lines. In this event they are less deep and are termed micro-die lines.
- (10) The origin of micro-die lines lies in the fracture and subsequent cavitation between fragments of coarse iron bearing precipitates when traversing the deformation zone, adjacent to the dead metal zone and on the die face. The cavitation appears as shallow score lines on the extrudate surface.

(11) The surface is generated from two deformation zones.

(a) the front of the main deformation zone, outlining the dead metal zone;

(b) a small deformation volume on the die face, adjacent to the die throat.

The latter deformation volume traps coarse precipitates which are subsequently released into the extrudate surface and along the die land area.

(12) Topological features can be incorporated in the limit diagram framework.

CHAPTER EIGHT

CHAPTER EIGHT

8.1. Concluding Discussion

The extrusion process is a convenient route for producing lengths of complex section which meet strict geometric, mechanical property and cosmetic goals. Because the AA6063 alloy is used for frames and decorative trim, the surface generating features are most important.

The use of an homogenisation treatment prior to extrusion improves the extrudability and the generated surface quality. This is primarily a consequence of the β - α iron phase transformation.

Extrusion of an homogenised AA6063 billet can be considered as a combined deformation and solutionising treatment, when suitably quenched from the hot deformation temperature. This is an important integral feature of commercial processing of AA6063, allowing subsequent ageing without the need for a separate solution heat treatment. In certain cases this has the added advantage of a contribution to tensile properties from the retained sub-structure.

The surface formation mechanism was identified as one of generation and deleterious modification. The process conditions for the production of an optimum surface quality were also identified. However, even at these conditions and when the die is operating in the choked condition, the elimination of the die line surface defect was not possible. It is likely that improvements can be made to reduce

the occurrence of die lines and micro-die lines further. However, this would be of academic interest only as such measures would be uneconomic in an already highly competitive market.

8.2. Recommendations for Further Work

- (1) The development of the mechanical equation state to include the minimum number of internal variables is essential, before it can be used to analyse and model the extrusion process. The equation should also be referred to a time base. This would allow inclusion of static softening features: clearly an important omission when used to analyse deformation schedules which involve numerous operations such as multi-pass rolling.
- (2) The use of Syalon or a similar 'non-wetting' die material should be used to examine the surface generating features. The use of such materials are reported to reduce pick-up, however their use in aluminium extrusion and the effect on the die line and micro-die line generation is yet to be established.
- (3) It would be useful to study the origin of micro-die lines further by using a super-purity based AlMgSi alloy. It is thought that extruding such an alloy at the optimum choked condition would produce an excellent finish due to the reduced iron content thereby reducing the percentage of deleterious iron phase.

- (4) An extension of the surface generation study to include more highly alloyed materials and RSP alloys would also assist in gaining further knowledge of surface generation.

APPENDIX ONE

Test Number	Billet (°C)	Temperatures	Extrusion	Ram Speed	lnZ	Peak Load	Peak Pressure	Strain-Rate
		T (°C)	Ratio	(mm ⁻¹)		(MN)	(MNm ⁻²)	(s ⁻¹)
106DC1-1d	550	-	10	3	21.4	0.67	152	2.1
106DC1-8d	450	-	10	3	24.3	1.13	256	2.1
106DC1-41d	250	50	10	9	34.2	2.85	643	5.9
106DC1-44d	500	-	10	9	23.6	1.09	246	5.0
381D-70d	450	-	20	50	27.5	1.20	271	30.0
381D-71d	450	-	20	50	27.5	1.28	289	30.0
381D-72d	450	-	20	50	27.5	1.16	262	30.0
381D-73d	450	-	20	50	27.5	1.26	284	30.0
381D-74d	450	-	20	50	27.5	1.26	284	30.0
DA18-51d	350	15	20	100	31.5	1.90	429	61.0
DA18-52d	300	33	20	100	33.8	2.39	539	61.0
DA18-53d	250	48	20	100	36.7	2.80	632	61.0
DA18-54d	200	59	20	100	40.0	3.18	695	61.0
381-168d	350	25	20	14	25.5	2.16	487	8.6
381-169d	550	-	20	14	22.9	0.93	209	8.6
381-170d	500	-	20	14	24.2	1.11	250	8.6
381-171d	450	-	20	14	25.7	1.37	309	8.6
381-172d	400	9	20	14	25.7	1.73	390	8.6
381-179d	450	-	20	14	25.7	1.31	296	8.6
381-180d	450	-	20	14	25.7	1.35	305	8.6
381-181d	450	-	20	14	25.7	1.34	302	8.6
381-182d	450	-	20	14	25.7	1.32	298	8.6
381-183d	450	-	20	14	25.7	1.34	302	8.6
381-184d	450	-	20	14	25.7	1.33	300	8.6
381-130i	450	-	20	14	26.1	1.42	320	12.9
381-131i	450	-	20	14	26.1	1.27	287	12.9
381-132i	500	-	20	14	24.6	0.96	217	12.9
381-137i	400	1	20	14	27.9	1.51	341	12.9
381-138i	350	13	20	14	29.9	1.85	417	12.9
381-149i	450	-	20	14	26.2	1.37	309	12.9
381-150i	450	-	20	14	26.1	1.30	293	12.9
381-151i	450	-	20	14	26.1	1.28	289	12.9
381-152i	450	-	20	14	26.1	1.32	298	12.9
381-153i	450	-	20	14	26.1	1.26	284	12.9
381-154i	450	-	20	14	26.1	1.29	291	12.9
HEW-202d	450	-	20	14	25.7	1.43	323	8.6
381-77d	450	1	30	35	25.7	1.49	336	8.6
381-82d	450	1	30	30	25.7	1.49	336	8.6
381-85d	450	2	30	30	25.7	1.54	347	8.6
381-87d	450	-	30	30	25.7	1.44	325	8.6
381-88d	450	1	30	30	25.7	1.50	338	8.6
381-89d	450	3	30	30	25.7	1.56	352	8.6
381-90d	450	1	30	30	25.7	1.51	340	8.6
106DC-28d	450	7	60	14	26.7	1.68	380	24.0
DA1-62d	450	100	60	8	26.4	4.18	943	17.0
DA1-63d	450	76	60	8	26.4	3.55	801	17.0
DA1-64d	450	7	60	14	26.7	1.66	375	24.0
381-79d	450	12	60	14	26.7	1.80	406	24.0
381-80d	450	12	60	14	26.7	1.81	408	24.0

381-81d	450	16	60	14	26.7	1.91	431	24.0
381-83d	450	16	60	14	26.7	1.91	431	24.0
381-84d	450	13	60	14	26.7	1.84	415	24.0
381-86d	450	15	60	14	26.7	1.89	426	24.0
381-105d	400	-	60	14	28.5	2.16	487	24.0
381-124d	480	58	60	14	25.8	3.07	693	24.0
381-125d	450	13	60	14	26.7	1.83	413	24.0
381-173d	500	2	60	14	25.2	1.53	345	24.0
381-174d	450	8	60	14	26.7	1.71	386	24.0
381-176d	400	29	60	14	28.5	2.28	514	24.0
381-177d	350	51	60	14	30.5	2.85	644	24.0
381-159d	550	-	60	14	23.9	1.25	282	24.0
381-160d	550	-	60	14	23.9	1.30	293	24.0
381-162d	550	-	80	14	24.2	1.30	293	32.9
381-163d	550	3	80	14	25.5	1.57	354	32.9
381-164d	450	16	80	14	27.1	1.91	431	32.9
381-165d	450	13	80	14	27.1	1.85	417	32.9
381-166d	400	31	80	14	28.8	2.33	526	32.9
381-167d	350	51	80	14	30.8	2.87	648	32.9
381-1281	450	8	450	14	27.1	1.71	386	32.9
381-1291	400	14	80	14	28.8	1.87	422	32.9
381-1341	500	-	80	14	25.5	1.45	327	32.9
381-1401	350	35	80	14	30.8	2.42	546	32.9
381-1331	500	-	40	14	24.8	1.25	282	16.0
381-1351	450	1	40	14	26.3	1.49	336	16.0
381-1361	400	10	40	14	28.1	1.74	393	16.0
381-1391	350	24	40	14	30.1	2.13	481	16.0
DA22-61d	450	8	60	14	26.7	1.70	384	24.0
381-92d	450	8	60	14	26.7	1.66	376	24.0
381-93d	450	18	60	14	26.7	1.96	442	24.0
381-95d	450	15	60	30	27.4	1.90	429	48.1
381-96d	450	14	60	60	28.1	1.86	420	95.5
381-97d	450	1	60	240	29.4	1.50	338	350.0
381-101d	400	32	60	14	28.5	2.35	530	24.0
381-114d	400	26	60	30	29.2	2.15	485	48.1
381-116d	550	-	60	14	23.9	1.26	284	24.0
381-117d	550	-	60	30	24.6	1.30	293	48.1
381-118d	550	-	60	60	25.3	1.19	262	95.5
381-119d	550	5	60	14	25.2	1.61	363	24.0
381-178d	350	51	60	14	30.5	2.86	644	24.0
381-196d	500	6	60	61	26.6	1.65	372	95.7
381-197d	400	26	60	61	29.9	2.20	496	95.7
381-198d	350	43	60	61	31.9	2.65	578	95.7
381-199d	440	32	60	34	29.3	2.35	530	54.1
381-200d	350	48	60	34	31.3	2.80	632	54.1
381-201d	500	5	60	32	26.0	1.61	363	51.3

APPENDIX TWO

APPENDIX TWO

Labmaster Data Collection

The experimental data for extrusion and torsion trials were collected using a PC-Mate Labmaster which converts the IBM PC into an Analog/digital and digital/analog converter. The Labmaster consists of a daughter board which performs the actual conversion of analog data to digital data; the information is then in a suitable form to signal to the computer, via a ribbon cable, to the mother board. Unlike the daughter board which resides outside the computer the mother board slots into the IBM PC bus. The Labmaster is a very versatile accessory and is described in full detail elsewhere⁽¹⁵⁾. In the following section the configurations used, and the relevant part of the control/collection program are documented.

IBM PC

The computer used for data collection had twin disk drives and expanded 256k memory. It also had a graphics card to enable high resolution display of the recorded data.

Mother Board

Residing in one of the expansion slots in the IBM bus, the mother board was selected to operate in the I/O (input/output) mapped mode where it responds to commands in the 8088 CPU. (This selection is

made using switches on the mother board.)(Figure A.1) The mother board requires 16 address locations which are necessary because of the multiple functions the board provides. There are numerous locations for the board to reside, but for the purpose of these trials the starting address recommended by Tecmar was used: that is 0170H (1808 decimal).

The 8088 CPU uses only an 8 bit data path and is only considered a 16 bit computer because within the IC it uses 16 bits. Therefore, when reading data to the computer it must be undertaken in two parts. First, the lower 8 bits are read, then the higher 8 bits. This sequence must be set into the programming. The reading of the higher 8 bits of data resets the Done and Overrun bits of the A/D. (An Overrun bit is set when the computer does not have quite enough time to move all the data to computer memory before the next data is available and so invalidates the data.

The mother board also carries an AMD 9513 timer which was used as a real time base for data collection.

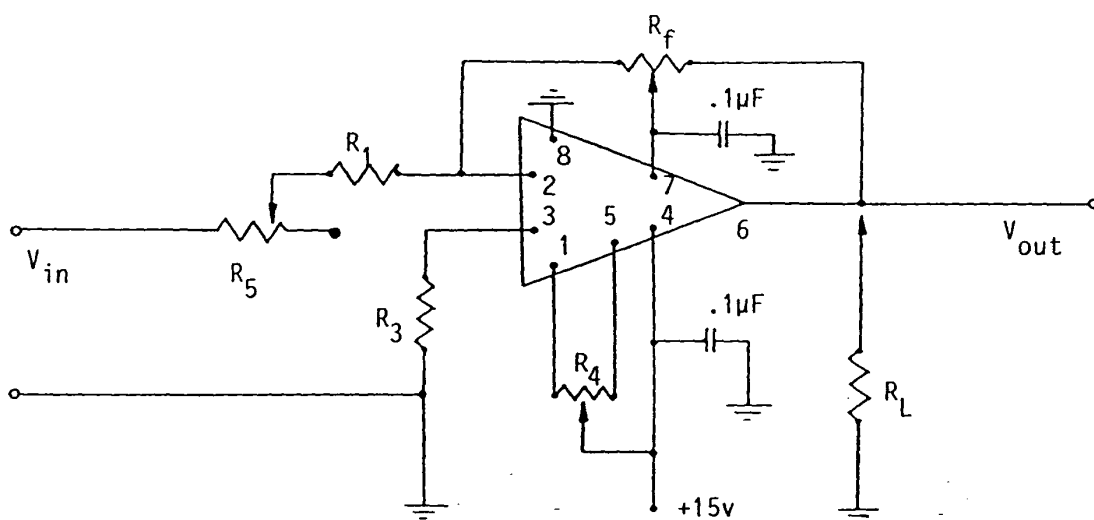
Daughter Board

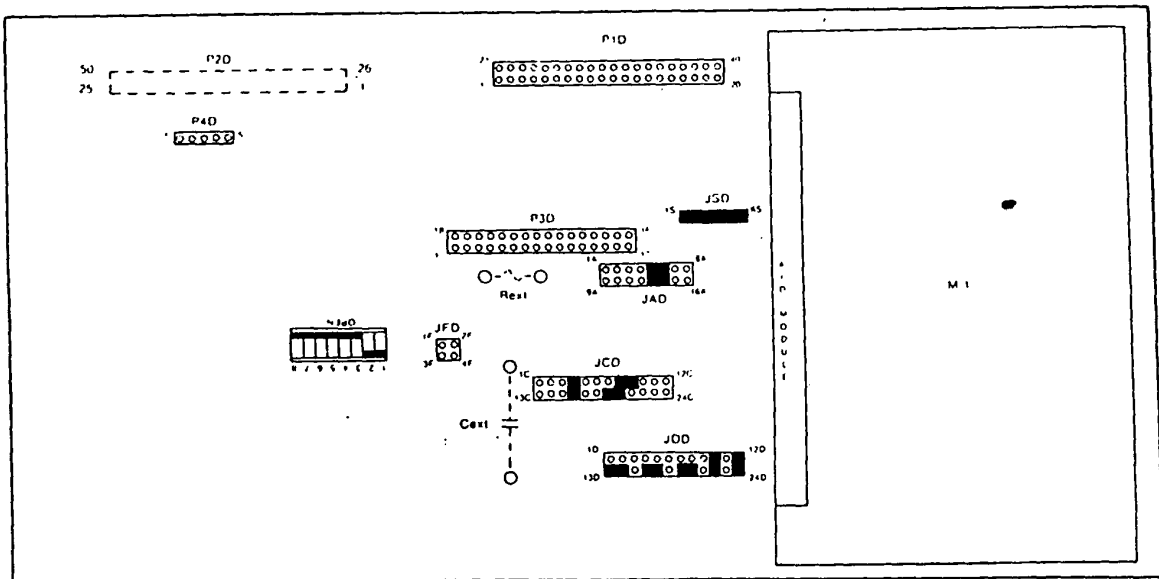
The major function of this board is to perform the actual conversion from analog to digital data. The data signals and control signals are all jumper selectable and are discussed in the following section.

The Labmaster allows, through programming, an automatic channel increment which permits up to 8 differential input channels to be

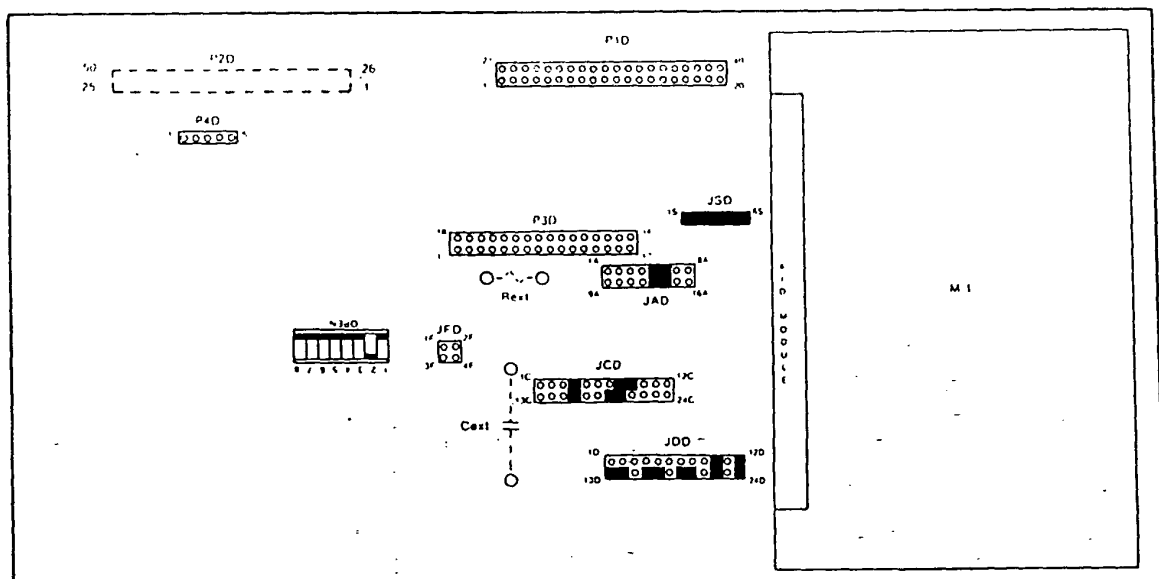
sampled. The starting channel is selected through software and the final channel through SW1D (Figure A.1) by selecting the correct switch configuration. (These are shown for both extrusion and torsion purposes.) The jumper selections made for both extrusion and torsion were common and are shown in Figure (A.1). These selections were for differential input, unipolar input in the +0 to +10 volts regime: The analog amplifications were performed on an 'in-house' designed op-amp circuit which is discussed in the next section.

OP-Amp Circuits





Jumper Selections and Switch Settings
For Torsion Data Acquisition.



Jumper Selections and Switch Settings
For Extrusion Data Acquisition.

Programming

The program steps necessary for operating the Labmaster in the required mode for these trials are described below: all programming was in BASIC.

```

1160 REM #####
1165 REM ##### SCAN LIMIT SWITCH ##
1170 REM #####
1175 OUT 1812,128:OUT 1813,6
1180 OUT 1814,0
1185 IF INP(1812)<128 THEN 1185
1190 L = INP(1813):HA = INP(1814)
1195 IF HA<5 THEN GOTO 1180

```

```

1200 REM #####
1201 REM ##### SET UP AUTO-INC. MODE ##
1202 REM #####
1205 OUT 1812,128:OUT 1813,255:OUT 1814,0
1210 IF INP(1812) < 128 THEN 1210
1215 X = INP(1814):OUT 1812,0:OUT 1813,0
1220 REM #####
1225 REM ##### SET UP COUNT 1 TOD MODE ##
1230 REM #####
1235 OUT 1817,23:OUT 1816,3:OUT 1816,129
1240 OUT 1817,1:OUT 1816,57:OUT 1816,31
1245 OUT 1816,0:OUT 1816,0:OUT 1817,65
1250 OUT 1817,9:OUT 1816,0:OUT 1816,0

```

```

1255 REM #####
1260 REM ##### DATA COLLECTION START ##
1265 REM #####
1270 DEF SEG = &H2000:OUT 1817,33
1275 FOR I = 0 TO 64000: STEP 10
1280 FOR J = 0 TO 6 STEP 2
1285 OUT 1814,0
1290 IF INP(1812) < 128 THEN 1290
1295 POKE(I+J),INP(1813):POKE(I+J+1),INP(1814)
1300 NEXT J
1305 OUT 1817,161:OUT 1817,17:POKE(I+8),INP(1816):POKE(I+9),INP(1816)
1310 IF PEEK(I+9) >= H THEN GOTO 1455

```

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