SIMULATION OF HOT WORKING OF AUSTENITIC STAINLESS STEELS

by

Ronaldo Antônio Neves Marques Barbosa, B.Eng., M.Sc.

Thesis submitted for the degree of Doctor of Philosophy.

University of Sheffield,
Department of Metallurgy,
December, 1983.

to Margaret and Simone

... Everyone who has done any kind of Creative work has experienced, in a greater or less degree, the state of mind in which, after long labour, truth or beauty appears, or seems to appear, in a sudden glory. It may be only about some small matter, or it may be about the universe. The experience is, at the moment, very convincing; doubt may come latter, but at the time there is utter certainty...

Bertrand Russell, 1946.

<u>CONTENTS</u>

SUMMARY

SIMULATION OF HOT WORKING OF STAINLESS STEEL

SUMMARY

The published literature on the strength and structural changes occurring during and after hot working of AISI316 and 304 austenitic stainless steels are reviewed.

Isothermal plane strain compression tests have been carried out with the purpose of determining relationships to describe the kinetics of static recrystallization, the recrystallized grain size, the isothermal grain growth rate and the strength during hot rolling of AISI316 steel. The kinetics of static recrystallization were also studied in samples tested in axisymmetric compression, or hot rolled. The effect on the kinetics of static recrystallization of the strain distribution in samples tested in plane strain compression was analysed.

The set of equations determined for 316 steel was used in a computer model modified from the one developed by Leduc (1980) for simulation of hot rolling loads and microstructural evolution. Partially recrystallized microstructure was generated in a laboratory hot rolling mill and was reasonably simulated by the use of the computer programme.

Non-isothermal plane strain compression tests were carried out for direct simulation of laboratory hot rolling results. Comparison between experimental hot rolling and

plane strain compression data has shown reasonable levels of agreement in the microstructural simulations undertaken in the present work. The mean plane strain strengths from non-isothermal plane strain compression tests were higher than the ones from hot rolling. This may have been caused by thermal gradients inside the sample being tested.

CHAPTER 1

Introduction

Hot working operations are those generally carried out at ~60% of the metal melting point and at strain rates within the range of 0.1 to 1000 s^{-1}. Most finished metallic products have been hot worked at some stage during their manufacturing route. The careful control of processing as well as material variables during hot rolling can be used to generate materials with better mechanical properties.

The simulation of hot rolling operations has been carried out in the past by numerical methods and by direct simulation using mechanical testing. Mechanical tests such as torsion or plane strain compression can provide information on the strength and structural behaviour of the material concerned. This information is translated into equations which can be used in a computer model to predict temperature, load and structural development during a given hot rolling schedule.

A number of hot workability tests have been used for hot rolling simulation. These are reviewed in chapter 4. Most simulations however, have been carried out under isothermal conditions. The few non-isothermal simulations which have been performed were not satisfactory since the cooling rate of the samples and interpass times differed

from the ones in the hot rolling process. Recently, non-isothermal hot rolling simulations were carried out in which the cooling rate of the specimen was controlled inside the test furnace so that it coincided with the one from the hot rolling slab. This technique allowed a direct comparison between microstructures with the specimen from the hot rolled slab. The hot rolling loads could also be suitably simulated (Foster,1981).

However, in most simulation work, the material entering the following deformation was either fully recrystallized, e.g C-Mn steels for instance in Leduc's work (1980) or fully deformed, e.g Nb-steels (Foster,1981); Al-1%Mn (Puchi,1983).

In the present work, complete hot rolling schedules have been undertaken on an experimental rolling mill using austenitic stainless steel type 316. This material presents partially recrystallized structures in between hot rolling passes, which permitted a more critical appraisal of the simulation techniques used.

It is expected that the methods employed in the present work may be applied to simulate industrial hot rolling schedules.

CHAPTER 2

Strength During Hot Deformation

2.1 Flow stress Relationships.

When stress-strain behaviour is examined over a wide range of different hot working conditions, it is usual to report a number of curves as typical examples and characteristic values of the flow stress as functions of temperature and strain rate. In order to correlate data at different temperatures, the Zener-Hollomon parameter is used, hence constant values of the activation energy, independent of the temperature, may be related to those for the rate-controlling process (Jonas et al.,1969). Sellars and Tegart(1972) have reviewed several equations which are attempts to fit simple algebraic functions to the stress-strain curve obtained at a constant temperature and strain rate. While recognizing the limitations of the expressions reviewed, they suggest that it is advantageous to apply them to experimental data so that constants can be reported from which flow stresses or the work done for any strain can be computed.

The steady-state data at low stress are best described by a

power relationship

$$\dot{\varepsilon} = A' \sigma_*^n \qquad (2.1)$$

where n is a temperature independent constant (Loizou and Sims,1953; Alder and Phillips,1954-55; Jonas et al.,1969). The steady-state data at high stresses are best described by an exponential relation

$$\dot{\varepsilon} = A'' \exp (\beta \sigma_*) \qquad (2.2)$$

where β is a temperature independent constant (Loizou and Sims,1953; Alder and Phillips,1954-55). The similarity between relations (2.1), (2.2) and the ones used for the steady-state creep led Sellars and Tegart(1966) to propose the expression

$$\dot{\varepsilon} = A (\sinh \alpha \sigma_*)^{n'} \exp (-Q / (R T)) \qquad (2.3)$$

where A, α and n' are temperature independent constants and Q is an activation energy. The constants α, β and n' are related by

$$\beta = \alpha n' \qquad (2.4)$$

At low stresses ($\alpha \sigma_* < 0.8$) equation (2.3) reduces to a power relation (equation (2.1)) and at high values of stresses ($\alpha \sigma_* > 1.2$) to an exponential relation (equation (2.2)). Figure(2.1) shows that equation (2.3) can be used successfully to correlate creep as well as hot rolling data for aluminium deformed under several different kinds of tests (Wong and Jonas,1968). The previous equations can be

used to predict a particular value of stress at a given
deformation condition. However, these relationships restrict
the application of the predicted value of stresses to strain
conditions equivalent to those under which σ_* was obtained.
Clearly, relationships capable of describing the stress at
any strain for a given temperature and strain rate are
required. For alloys which present a peak stress,
Leduc(1980) has used the following set of equations, which
apply to C-Mn steels:

$$\sigma = \sigma_e \qquad \text{if} \qquad \varepsilon < (x \ \varepsilon_p) \tag{2.5}$$

$$\sigma = \sigma_e - \Delta\sigma \quad \text{if} \qquad \varepsilon > (x \ \varepsilon_p) \tag{2.6}$$

where

$$\sigma_e = \sigma_o + B \ (1 - \exp - C \ \varepsilon \ )^m \tag{2.7}$$

(Sah et al., 1969)

$$\Delta\sigma = B_1 \left[1 - \exp \left[- k \ \frac{\varepsilon - x \ \varepsilon_p}{\varepsilon_p} \right]^{m'} \right] \tag{2.8}$$

where

$$B = \sigma_{ss(e)} - \sigma_o \tag{2.9}$$

$$B_1 = \sigma_{ss(e)} - \sigma_{ss} \tag{2.10}$$

$$C = - 10 \ln \left[1 - \left[\frac{\sigma_{0.1} - \sigma_o}{B} \right]^{1/m} \right] \qquad (2.11)$$

The σ subscripts o,ss,ss(e) stand for zero strain, steady-state and steady-state extrapolated using equation (2.7). ε_p is the peak strain, n~0.8 for C-Mn steels and n, m, m' are temperature independent constants. Equations (2.5) and (2.6) give a complete description of the stress-strain curve provided that the original grain size and deformation conditions are known. Figure(2.2) illustrates the computational procedure used to calculate a stress-strain curve for C-Mn steel. At the same time an experimental curve is shown for comparison purposes.

2.2 Strength During Hot Working of Austenitic Stainless Steels.

The maximum hot stress, σ_p, and the stress to a strain equal to 0.1, $\sigma_{0.1}$, obtained from several different sources for austenitic stainless steels type 316 and 304 hot worked under different kinds of testing conditions, with temperatures ranging from 800 to 1200 C and strain rates no higher than $100s^{-1}$ but mostly less than $5s^{-1}$ are shown in tables (2.1) and (2.2). It can be seen from table(2.1) that Q_{def}, the activation energy for deformation, varies from author to author although the 3 reported values are for

materials with nearly the same chemical composition and tested under hot torsion. It can be written then that

$$\bar{Q}_{def} = 456 \quad kJ/mol$$

where the bar stands for average. The reported values of Q_{def} for 304 type steel are slightly higher for torsion than the ones calculated from axisymmetric compression tests. The average value of Q_{def} for 304 steel being

$$\bar{Q}_{def} = 405 \quad kJ/mol$$

therefore it may be inferred that

$$\bar{Q}_{def316} = 1.12 \quad \bar{Q}_{def304}$$

It is interesting to note that the strengths of 316 and 304 steels are in similar proportion to their Q_{def} values. Figure(2.3) and (2.4) show the dependence of σ_p and $\sigma_{0.1}$ for the 316 steel on the Zener-Hollomon paramenter. It seems that an exponential type of equation describes reasonably the stress dependence on Z for all levels of stresses reviewed. As expected, the scatter obtained when $\sigma_{0.1}$ is plotted is bigger than that obtained for σ_p. This is simply explained in terms of the steepness of the stress-strain curve in the work hardening region. It is interesting to note that σ_p is approximately equal to $\sigma_{0.1}$ at Z values of $\sim10^7$ (T$\sim$1200 C at $\dot{\varepsilon}$ $\sim$5s^{-1}). σ_p then becomes increasingly higher than $\sigma_{0.1}$ as temperature decreases. The same characteristics shown by the 316 steel can be seen in the 304 type as depicted in figures (2.5) and (2.6).

Figures (2.7a) and (2.7b) compare levels of stress ($\sigma_p, \sigma_{0.1}$) for both steels when hot worked at a constant strain rate ($5s^{-1}$) at several temperatures, whereas in figures (2.8a) and (2.8b) the temperature is kept constant (~1100 C) and the strain rate is varied within the usual strain rate range for hot working operations. As it can be seen, the stress level for AISI316 steel is ~1.15 higher than the ones for 304 type steel hot worked under the same conditions in agreement with previous reports (Gittins et al.,1977; Bywater and Gladman,1976). This higher strength is attributed to the solid solution strengthening of molybdenum (Bywater and Gladman,1976).

Figures (2.9) and (2.10) show the correlation of the strain to the peak stress with the Zener-Hollomon parameter for 316 and 304 steels. The first feature to be observed is that there is some scatter for the set of data presently available for both steels, even when a single author is considered for analysis. This may be explained in terms of the shape of the stress-strain curves for 316 and 304 steels. Usually, at strains near to the peak strain, the stress-strain curve for austenitic stainless steels presents approximately null work hardening rate for a wide band of strains, leading to uncertainty in determining the ε_p value. It has been reported (Leduc,1980) that for strain rates higher than $5s^{-1}$, an increase in the strain rate leads to a decrease in the ε_p value. Most of the data available from tables(2.1) and (2.2) are for strain lower than $5s^{-1}$. However, the trend reported by Leduc (1980) could only be

observed for plane strain compression samples (Colàs,1983). On the other hand, for other types of tests, ε_p increased very gently with Z as seen in figures(2.9) and (2.10). There is some uncertainty about the effect that the original grain size might have on the maximum flow stress. Leduc's model (1980) for predicting stress-strain curves for C-Mn steels has shown that increasing the initial grain size increases the maximum strength significantly. This arises because of the strong dependence of ε_p on the original grain size for C-Mn steels (Sellars,1980). Foster(1981) has reported an opposite effect on σ_p for Nb-steels while Roberts et al. (1979) have found no effect of d_o on the maximum strength value. However, the latter used materials with 75 to 200 μm original grain size, whereas the former used samples with d_o ranging from 9 to 94 μm.

2.3 Flow Stress and Strain Localization.

The plane strain compression test has been widely used as a method of obtaining values of the flow stress of materials, so that the metal strength properties appropriate to the hot working operations can be determined. Sellars et al.(1976) testing lead under plane strain compression have shown that a series of stress-strain curves when properly corrected for the effects of spread, friction and temperature rise as a result of deformational work showed a

considerable effect due to differences in the initial sample geometry, in particular before the peak stress. Aluminium specimens tested at room temperature and partially recrystallized by annealing presented evidence of heavily deformed regions, revealed by bands of recrystallization, and less deformed zones, shown by unrecrystallized material (Beynon,1979; Puchi,1983; Beynon and Sellars,1983). The differences in the stress-strain curves at high temperature and the metallographic evidence suggested that they might have arisen from differences in strain distribution in samples of different original geometries.

A detailed investigation in gridded aluminium (Beynon,1979; Beynon and Sellars,1983) led to the conclusion that the differences in the stress-strain curves were due to different amounts of redundant work involved in deforming samples of different starting geometries. An empirical correction for this effect was proposed and corrections for Al-1%Mg-1%Mn deformed under plane strain compression proved to be quite satisfactory (Beynon,1979). However, although the technique improved agreement between curves from different geometries for Nb-steels, Foster(1980) acknowledged that differences between curves remained still significant.

Experimental evidence for the existence of shear bands has been reported in the literature (Amiya et al., 1975a; Semiatin and Lahoti,1981a,81b) and a criterion for plastic instability occurring by shear has been formulated (Amyia et al.,1975b). Studying non-isothermal forgings, Semiatin and

Lahoti(1983a) determined conditions of material properties and process parameters which led to shear bands in the material being processed. They used Ti-6242 with two different microstructures: $\alpha + \beta$ microstructure (an equiaxed α grain structure in a transformed β matrix) which does not exhibit flow softening during test and β microstructure samples (an acicular transformed microstructure) which soften during flow. They concluded that the flow localization is greater in the β structure than the one found in the $\alpha + \beta$ samples. On the other hand, there have been suggestions that flow localization (and the consequent drop in the stress-strain curve) is a phenomenon which is solely caused by strain localization to the dynamic bands rather than as a result of cyclic recrystallization (Korbel and Blaz,1980).

Evaluations of mechanical properties from forged samples containing and not containing shear bands demonstrated that the effect is small, if not negligible, on the tensile and fatigue loads at ˜500 C for Ti-6242 alloy, although an improvement in the stress-rupture properties has been reported (Semiatin and Lahoti, 1983a,83b). However, when deformed in torsion at temperatures ranging from 816 to 1010 C, $\alpha + \beta$ material consistently showed ϵ_p values greater than the ones found for β material. This has been mainly attributed to changes in the original, non-equilibrium, microstructure including Widmanstätten morphology and dislocation substructure (Semiatin and Lahoti, 1981a,81b).

The deformation in plane strain has been simulated using

numerical methods (Tvergaard et al.,1981). They suggested that the localization of shear bands in the sample being deformed is strongly influenced by small strain concentrations near the surface of the sample. Moreover, the intensity of deformation varies along the bands and some of the shear bands end inside the material.

From all that has been mentioned above, it appears clear that the effect that concentration of strain in shear bands might have on the flow curve during hot working still remains not fully unveiled.

CHAPTER 3

Structural Changes During and After Hot Working

3.1 Introduction.

Hot working operations are those carried out at high temperature (>0.6 T_m, where T_m is the melting temperature in degrees Kelvin) and relatively high strain rates (10^{-1} to $10^3 s^{-1}$) (Sellars and Tegart,1972). The material being processed may undergo dynamic changes (during deformation) as well as static changes (during rest periods) in its structure. Dynamic recovery can sometimes be the only softening process occurring during deformation. However, in austenitic stainless steels dynamic recrystallization could take place at some stage in a typical hot rolling operation schedule. As soon as deformation ceases, further microstructural changes (static recovery, recrystallization and grain growth) will determine the structure of the material entering the following pass.

A number of reviews have been published in the literature on the subject (Jonas et al.,1969; McQueen and Jonas,1975). In the present chapter a brief outline of the physical

metallurgy of hot working is given. Particular emphasis is placed on the basic information needed for a hot rolling simulation to be carried out. Since we are concerned with hot rolling operations of austenitic stainless steels, data available in the literature on AISI316 and AISI304 steels are reviewed and analysed.

3.2 Dynamic Recovery.

Dynamic recovery is the term used to describe the thermally activated recovery occurring during deformation at high temperatures. It is always present during the deformation process and in high stacking fault energy materials such as aluminium, α-Fe and ferritic alloys it may be the only dynamic softening mechanism. The flow stress rises during the early working hardening stage, in which a tangled dislocation substructure is generated. By the time the steady-state is achieved, the dislocations have arranged themselves into subgrains whose perfection, dimensions and misorientation depend on the metal, on the deforming strain rate and temperature (Jonas et al.,1969). During steady-state, an equilibrium between the rate of dislocation generation and annihilation is achieved. This gives an equiaxed appearance to the subgrain structure and a constant subgrain size during the steady-state regardless of the strain applied. On the other hand, the grains deform in conformity with the outward change in shape of the sample.

This cannot be explained solely by migration of sub-boundaries (McQueen and Jonas,1975). Another contributing process is repolygonization, i.e., the repeated unravelling of the sub-boundaries and subsequent formation of new sub-boundaries at locations which keep their average spacing and dislocation density constant (McQueen et al.,1967).

The dependence of the mean subgrain diameter d_s on temperature T and strain rate $\dot{\varepsilon}$ is described by the relationship (McQueen and Jonas,1975)

$$d_s^{-1} = a + b \log_{10} Z \tag{3.1}$$

where a,b are constants and Z is the Zener-Hollomon parameter. The mean subgrain size during the steady-state is uniquely related to the steady-state stress by

$$\sigma_{ss} = k \, d^{-1} \tag{3.2}$$

where k is an empirical constant.

3.3 Dynamic Recrystallization.

Metals with low stacking fault energy (copper, nickel, austenitic steels) have a poorly developed substructure formed by the slow recovery process during the early stages of deformation. When a critical density of dislocation is achieved inside the grains, nucleation and dynamic recrystallization takes place. Figure(3.1) shows schematically the effects that dynamic recrystallization may

have on the stress-strain curve of the material being deformed. At a low strain rate ($\dot{\varepsilon} < 0.1 s^{-1}$) the curve presents several localized maxima which decrease in amplitude until a steady-state stress flow level is achieved, whereas at high strain rates only one maximum value of stress is found. There have been several reviews on the subject of dynamic recrystallization in the past few years (Jonas et al.,1969; Sellars and Tegart,1972; McQueen and Jonas,1975; Sellars,1978, 1979, 1981a).

In the summary that follows, the main findings to date on the phenomenon of dynamic recrystallization are presented. More attention is given to points which are thought to be of direct interest for hot rolling simulation:

a- It is generally recognized that a critical strain, ε_c, has to be achieved in order for nucleation of dynamic recrystallization to take place. ε_c can be written as:

$$\varepsilon_c = a \, \varepsilon_p \qquad\qquad (3.3)$$

where "a" has been reported to be 0.86, 0.83, 0.67 (Sellars and Tegart,1966; McQueen and Jonas,1975; Sellars,1980). The work hardening rate of the flow stress curve in the vicinity of the maximum stress tends towards zero, therefore $\sigma_c = \sigma_p$.

b- The critical strain, ε_c, decreases monotonically with the Zener-Hollomon parameter until a minimum value is achieved at low stress values (Richardson et al. ,1966). From there on, there is an increase in the ε_c

for a decrease in the stress. This is due mainly to the fact that a low dislocation density exists because the applied stresses are very low, therefore a higher value of strain is expected to be necessary to produce a critical dislocation density for recrystallization (Sellars, 1978).

c- The dynamically recrystallized grain size(d) increases with the decrease in maximum stress, σ_p, and the steady-state stress, σ_{ss}. It can be written that

$$\sigma_{ss} = \sigma_o + A \ d^{-n} \qquad (3.4)$$

where σ_o, A and n are empirical constants with n having a value in the range of 0.5 to 1.0 for several metals (Sellars, 1979) and $\sigma_o \sim 0$ (Luton and Sellars, 1969).

d- The original grain size (d_o) affects the rate of dynamic recrystallization (Roberts et al., 1979). As the original grain size becomes coarser, it causes a decrease in the rate of recrystallization. The kinetics data can be described by an Avrami law where the time exponent k is approximately equal to 1.3 for austenitic stainless steels.

e- The dynamically recrystallized grain size is smaller than the statically recrystallized one, as it can be seen from figure (3.2) (Ahlblom, 1977a).

A phenomenological model to describe the stress-strain behaviour has been proposed (Sellars, 1969; Luton and Sellars, 1969) and has been experimentally checked (Sah et al., 1969; 1973). Formal theories of dynamic recrystallization

based on dislocation density changes and grain boundary migration kinetics have been put forward by Stüwe and Ortner(1974). A criterion for dynamic recrystallization proposed by Sandström and Lagneborg (1975a) takes into account the existence of a gradient of dislocation density inside the nuclei. This was subsequently improved (Roberts and Ahlblom,1977) by considering that the reduced driving force within the nuclei should be related to a radial average of the dislocation density difference and not to the minimum dislocation density difference within the potential nucleus (Sandström and Lagneborg,1975b).

3.4 Metadynamic Recrystallization.

Whenever a material being processed undergoes strains higher than the peak strain during hot working operations, metadynamic recrystallization will to some extent be responsible for the softening of the material during rest periods. If the prior strain imparted to the sample is less than ε_{ss}, the strain for steady-state to be achieved, there will be static recovery and static recrystallization in addition to metadynamic recrystallization as softening mechanisms. However, at strains higher than ε_{ss}, static recovery and metadynamic recrystallization will be the only softening processes to occur (see figure(3.3)). Unlike its static counterpart, metadynamic recrystallization does not exhibit an incubation

time, indicating that nucleation takes place during deformation (Djaic and Jonas,1972). It has also been reported (Ahlblom,1977a) that metadynamically recrystallized grain sizes are finer than the static recrystallized ones and slightly coarser than the dynamically recrystallized ones, as it can be seen from figure(3.2).

3.5 Static Recovery.

Static recovery occurs whenever deformation is halted during hot working processes and it may be the only softening mechanism present when strains less than the critical ones for static recrystallization (generally less than 10%) are applied to the sample (McQueen and Jonas,1975; Djaic and Jonas,1973). In cases in which static recovery is the only mechanism involved, softening cannot reach 100%; values of softening being reported to be ~40 to 60% (Djaic and Jonas,1972). The rate of static recovery increases with temperature, strain and strain rate and decreases with the increase of alloying additions (Petkovic et al.,1975).

It has been recognized (Barraclough and Sellars,1979) that continuing effects of static recovery on static recrystallization may be deemed responsible for a non-linear correlation between static softening measured through restoration index, such as the yield stress or hardness, and static softening followed by metallographic techniques. At the same time Barraclough and Sellars(1979) acknowledge that

non-homogeneous distribution of strain throughout the sample may also have caused the non-linearity in the above mentioned correlation. It has been suggested (Towle and Gladman,1979) that a decrease in the rate of static recrystallization may be explained in terms of a decrease in the driving force for static recrystallization due to softening by static recovery occurring during the incubation period of static recrystallization.

3.6 Static Recrystallization.

3.6.1 Kinetics of Static Recrystallization.

The general rate laws of transformation associated with the static recrystallization of metals were first derived by Johnson and Mehl (1939). They assumed that the nucleation process occurred at random and that the transformed product could not grow across grain boundaries. Experimental data related to kinetics of static recrystallization are commonly plotted according to the Avrami equation (Avrami,1941)

$$X = 1-\exp(-\beta t^k) \qquad\qquad (3.5)$$

Where X is the volume fraction recrystallized, t is the period of time in which the sample was held at an annealing temperature, and β and k are empirical constants at a

constant temperature and prior strain. The value of k is determined by details of the nucleation and grain growth processes while the constant β additionally depends on the driving force for recrystallization and the migration velocity of the grain boundaries (Ahlblom,1977a). Cahn(1956) treated the kinetics of grain boundary nucleated reactions by assuming that:

a- the growth rate and the specific nucleation rate are independent of time after steady-state conditions are achieved.

b- grain boundaries offer no resistance to transformation.

c- nucleation occurs randomly.

d- only one kind of site is active in nucleating.

He then showed that the k value in the Avrami equation is expected to be ~4 during the early stages of the reaction, i.e., before site saturation happens. When site saturation occurs, there is a change in k value to 1,2 or 4 according to the nucleation sites being grain boundaries, grain edges or grain corners. The value of β and k however have been reported to be dependent on prior strain applied to the material (Wusatowski,1966). Indeed, it has been shown that a k value of 1 is expected when α-iron is deformed so that dynamic recrystallization preceeds static recrystallization, whereas the k value is expected to be 2 for materials that are only dynamically recovered (Glover and Sellars,1972). More recently, k values less than 2 have been reported for austenitic stainless steels (Towle and Gladman,1979; Al-

Jouni,1983). This is usually attributed to the fact that the growth rate during transformation is not constant with time. Wittaker(1973) found values of k~2 for 3% silicon steel deformed in torsion, although the nucleation process was clearly carried out in the grain boundaries and the growth rate during recrystallization was not constant with time. Static recrystallization data on AISI316 and AISI304 from several authors are reviewed and tabulated as shown in tables (3.1) and (3.2). It can be seen from those tables that k values can be distributed as follows:

Steel	Total Number of observations	% of total no. of observations			
		>2.0	2.0<k<1.5	1.5<k<1.0	<1.0
316	22	9	26	52	13
304	38	23	15	25	35

Most of the k values found are less than 1.5, which confirms reports of grain boundary as the main nucleation site. There is no indication that the k value would vary with the type of the test for AISI316, whereas for 304 type steel a higher value of k (towards 2) has been found for torsion (Barraclough and Sellars,1979; Lombry et al.,1980). It is difficult to analyse from the present set of data the

influence that testing as well as material variables may have on the value of k. However, it seems that for rolling at least, it has been suggested that k increases as the applied strain increases (Towle and Gladman,1979). As far as the strain effect on k value is concerned, the same general trend is confirmed by Ahlblom(1977a) for AISI304, whereas the opposite effect was found by Nörstrom(1977), confirming findings by Campbell(1974).

3.6.2 Activation Energy for Static Recrystallization.

Nucleation and grain growth are two different processes which occur separately and simultaneously during static recrystallization of a deformed metal. Each of these stages can be described by an Ahrrenius type of expression with an activation energy associated to it. The activation energy for nucleation and the one for grain growth are dependent upon prior strain applied to the material as can be seen from figure(3.4) (Anderson and Mehl,1945). An overall activation energy for the process of static recrystallization is a function of the activation energy for nucleation and for grain growth. It has been suggested (Cotteril and Mould,1976) that calculations of an 'overall activation energy' for recrystallization (based on the rate of change of some physical property, or on the recrystallized fraction) have little real value, since the overall activation energy is a complex function of the activation energies for two individual processes. However,

site saturation is achieved early during the recrystallization process (Cahn,1956). Additionally, it has been verified that although the recrystallization rate is theoretically governed by both the nucleation and the growth rates, the recrystallization rate becomes a function of the growth rate and independent of the nucleation rate, provided intragranular nucleation is insignificant and that sufficient new grains nucleate in each old grain early during the reaction (Cahn and Hagel,1962). Therefore, it may be excusable to use parameters such as t_{50}, time for 50% of material to recrystallize, to calculate an activation energy for the process. Whittaker(1973) proposes that an activation energy for static recrystallization can be evaluated by

$$Q_{rex} = -R \left[\frac{\delta \ln 1/t_{50}}{\delta 1/T} \right]_{\sigma_f, \epsilon_f} \tag{3.6}$$

where σ_f is the finishing stress and ϵ_f is the finishing strain other symbols having their usual meaning. This approach implies that the driving force for static recrystallization is a sole function of Z, the Zener-Hollomon parameter. However, doubts have recently been cast over the dependence of the stored energy on Z (Ahlblom and Sandström,1982). Figure(3.5) shows the dependence of the flow stress on recrystallization rate (measured as the difference between t_{60} and t_{40}) (Ahlblom,1977a; Ahlblom and Sandström,1982). The authors suggest that σ_f is not a good measure of the driving force for static recrystallization. This is clearly seen when σ_f changes are induced by

variations in the strain or strain rate at a constant temperature. The authors concluded that the strain rather than the finishing stress seems to be the essential parameter determining the driving force for static recrystallization (Ahlblom,1977a; Ahlblom and Sandström, 1982).

Several values of activation energy for static recrystallization are reported in tables (3.1) and (3.2) for AISI316 and AISI304 respectively. It can be written that

$$\bar{Q}_{316} = 431 \quad kJ/mol$$

and

$$\bar{Q}_{304} = 366 \quad kJ/mol$$

where $\bar{Q}$ stands for average activation energy. Therefore

$$\bar{Q}_{316} = 1.18 \ \bar{Q}_{304}$$

The fact that the activation energy for static recrystallization for AISI316 is higher than the one for AISI304 is usually attributed to being caused by the presence of molybdenum in the 316 type steel (Towle and Gladman,1979).

3.6.3 Parametric Equations on Recrystallization Rate.

Morrison(1972) proposed that the strain dependence on the t_{50} can be written as :

$$t_{50} = C_{0.5} \; \varepsilon^{-p} \; \exp\frac{Q_{rex}}{RT} \tag{3.7}$$

where $C_{0.5}$ is a constant for a given grain size and showing no dependence on the strain rate. Therefore, t_{50} can be said to be solely dependent on strain, p being equal to -4 in the case of low carbon steels in the austenitic range. Barraclough extended equation (3.7) to include the effects of the original grain size, d_o, the strain rate and the temperature through the use of the Zener-Hollomon parameter (Barraclough,1974) as follows:

$$t_{50} = C_{0.5}\varepsilon^{-p} \; \exp\frac{Q_{rex}}{R\,T} \; A \; Z^{-q} \; d_o^n \tag{3.8}$$

where p=3.5, q=0.38 whereas n is a function of strain, decreasing as the strain increases. Sellars and Whiteman, after reviewing several sets of published data in the literature on C-Mn steels (Sellars and Whiteman,1976), came to the conclusion that:

$$t_{50} = A' \; d_o^2 \; \varepsilon^{-4} \; Z^{-0.375} \; \exp\frac{Q_{rex}}{R\,T} \tag{3.9}$$

They suggested that there is an overall agreement as far as the strain and the Zener-Hollomon parameter exponents are concerned. At the same time, they recognize that the original grain size dependence is somewhat uncertain. A more recent review on C-Mn, low-alloy and Nb-treated HSLA steels

(Sellars,1980) has shown that in fact the t_{50} equation may be split into at least two different relationships according to the prior strain applied to the sample. This can be seen in figure (3.6) for C-Mn and low-alloy steels. The equations proposed are of the type

$$t_{50} = B \; \varepsilon^{-4} \; d_o^2 \; \exp\left[\frac{300000}{R\,T}\right] \qquad \text{when} \qquad \varepsilon < 0.8\varepsilon_p \qquad (3.10)$$

and

$$t_{50} = B' \; Z^{-0.6} \; \exp\left[\frac{300000}{R\,T}\right] \qquad \text{when} \qquad \varepsilon > 0.8\varepsilon_p \qquad (3.11)$$

where ε_p is the peak strain. Equations (3.10) and (3.11) confirm findings (English and Backofen,1964; Glover and Sellars,1972; Morrison,1972; Djaic and Jonas,1973) that the strain exponent is -4 and that t_{50} is independent of the strain after a certain critical strain ε_c is achieved. The activation energy is an average value for several steels analysed and is assumed to be constant with strain. Towle and Gladman (1979) found a dependence of t_{50} on d_o lower than an exponent of 2 for both 304 and 316 steels.

An overall relationship for all values of strain has been proposed (Barraclough and Sellars,1979)

$$t_{50} = A \; Z^{-0.38} \; d_o^2 \; \exp\left[\frac{\dfrac{Q_{rex}}{R\,T}}{1-\exp-\;1.5\left[\dfrac{\varepsilon-0.08}{\varepsilon_p}\right]^2}\right] \qquad (3.12)$$

However, the constant d_o exponent causes t_{50} to be a function of the original grain size regardless of the strain level applied to the sample.

Data on 304 and 316 type steels are analysed according to

$$B \; t_{50} \; = \; A \; exp\frac{Q_{rex}}{R \; T} \qquad (3.13)$$

where

$$B \; = \; \epsilon^4 \; \frac{Z^{0.38}}{d_o^2} \qquad (3.14)$$

Figures (3.7) and (3.8) show the result obtained from data on 316 and 304 steel respectively. It can be seen that:

a- The overall agreement is reasonable for both steels (within two orders of magnitude for the overall data set if the rolling data are excluded) supporting the idea that equation (3.13) may be of a reasonable form for describing the recrystallization rate of austenitic stainless steels. However, figure(3.7-b) shows that a strain exponent of 3 may yield a slightly better correlation than a fourth power. A change in the d_o exponent from 2 to 1 does not alter significantly the present scatter. The activation energy for recrystallization varied from ~450 to ~560 kJ/mol for 316 type steel and from ~360 to 460 kJ/mol for AISI304. The differences between authors seem to originate more from the different kinds of tests used for data acquisition rather than from differences in chemical compositions.

b- An increase in the molybdenum content from ~2.5%wt to 4%wt (nominally AISI310 steel) does not seem to alter the rate of recrystallization significantly.

c- Specimens from hot rolling samples gave longer t_{50} values than the ones found when using other testing methods. This may be due to the fact that the temperature of the sample drops when a rolling pass is given and a finite time interval is needed for the specimen to regain the annealing temperature when it is put back into the furnace.

d- The recrystallization rate for both steels seems to be very similar at high temperatures (1/T~6.5 to $7.5 \times 10^{-4} K^{-1}$ or T~1250 to 1050 C). However, they differ considerably at lower temperature (< 1000 C). This may be explained by the higher activation energy for the 316 type steel.

3.7 Recrystallized Grain Size.

Phenomenological equations similar in form to those used to describe the rate of static recrystallization were first proposed by Sellars and Whiteman (1976) after reviewing published data on C-Mn available in the literature. They came to the expression

$$d_{rex} = 25 \; \varepsilon^{-1} \; d_o^{0.5} \left[\frac{1}{\beta} \ln \frac{Z}{A} \right]^{-0.67} \qquad (3.15)$$

where A and β are experimentally determined constants in the

relationship

$$Z = A \exp \beta \sigma \qquad (3.16)$$

They suggested that the constant d_o exponent equal to 0.5 might have similar limitations to the one for t_{50}, equation (3.9). The recrystallized grain size becomes independent of the strain for strains higher than ϵ_*, critical strain, which is $\sim 2\epsilon_p$ for α-iron (Glover and Sellars, 1972). Equation (3.15) may only be applicable for strains less than ϵ_*. A thorough review on C-Mn, low alloy and Nb-treated HSLA steel data (Sellars, 1980) led to

$$d_{rex} = D \, d_o^{0.67} \, \epsilon^{-1} \qquad (C\text{-}Mn, \epsilon < \epsilon_*) \qquad (3.17)$$

$$d_{rex} = D' \, d_o^{0.67} \, \epsilon^{-0.67} \qquad (Nb, \epsilon < \epsilon_*, T > 950C) \qquad (3.18)$$

$$d_{rex} = D_* Z^{-n} \qquad (C\text{-}Mn, \epsilon > \epsilon_*) \qquad (3.19)$$

where D, D' and D_* are constants. Additionally,

$$D_* = k \, D \qquad (3.20)$$

where k is a constant, and

$$\epsilon_* = (0.57 \, d_o^{0.17}) \, \epsilon_p \qquad (3.21)$$

It has been attempted to describe d_{rex} in terms of a single expression. Barraclough and Sellars(1979) studying austenitic stainless steels type 304 came to

$$d_{rex} = B \ Z^{-0.1} \ \frac{d_o^{0.5}}{1.15 - \exp - 2.5 \frac{\varepsilon - 0.08}{\varepsilon_p}} \qquad (3.22)$$

The exponential term leads to a quick decrease in the effect of the strain on d_{rex}, after a strain higher than ε_p is applied to the sample. However, d_{rex} remains dependent on the original grain size at any strain, contradicting findings for C-Mn and HSLA steels (Sellars,1980). Towle and Gladman (1979) successfully correlated experimental data for AISI316 by using

$$d_{rex} \ \alpha \ \varepsilon^{-0.5} \ d_o \ Z^{-0.06} \qquad (3.23)$$

In order to correlate experimental data on both 316 and 304 type austenitic stainless steels obtained from the literature and given in tables (3.3) and (3.4), the relationship

$$d_{rex} \ \alpha \ \varepsilon^{-1} \ d_o^{0.5} \ Z^{-0.06} \qquad (3.24)$$

has been used.

Figures (3.9) and (3.10) show the degree of correlation obtained when using equation (3.24) especially if groups of different kinds of tests are separately analyzed. The agreement obtained may confirm that equation (3.24) is at least of the correct form for strains less than or approximately equal to the peak strain. Although it is recognized that there are only a few experimental points reported on 316 steel, it may be seen from figure (3.9) that changes in the molybdenum content may have little effect on

the recrystallized grain size.

Figure (3.11) shows the dependence of the recrystallized grain size on the equivalent strain for both 316 and 304 type steels. It can be seen from figure (3.11) that a slope ~1 is obtained, supporting the use of $d_{rex} \alpha \epsilon^{-1}$ in equation(3.24).

It can also be noticed that the recrystallized grain sizes for both kind of steels are very similar, although, at the same time, there is an increasing experimental scatter for strains ~0.1. This maybe so because at these strains d_{rex} becomes more strongly dependent on strain (~-4 exponent).

It is not possible from the data available to decide if there is a sharp or a smooth transition between the two different regimes.

3.3 Grain Growth.

The process of static recrystallization is considered to have finished when the boundaries of the newly created, strain-free grains have migrated through the sample to such an extent that they impinge on each other. After static recrystallization is over, the grains will continue to grow further in order to reduce the energy per unit volume associated with grain boundaries. Empirical relationships to describe the grain growth occurring during isothermal annealing were first put forward by Beck et al.(1948). He concluded that normal isothermal grain growth

could be described by a parabolic equation

$$d = k\ t^n \tag{3.24}$$

Where d is the average grain size after an annealing period t, and k and n are constants dependent on the composition of the metal and annealing temperature but independent of the grain size. The same author however, proposed a more generalized form of equation (3.24) as

$$d^{1/n} - d_{rex}^{1/n} = kt \tag{3.25}$$

where d_{rex} is the recrystallized grain size and n and k are constants (Beck, 1948).

Derivation of relationships describing grain growth from basic principles came to an equation of the same form as equation (3.25) but with a grain size exponent of 0.5. Burke's analysis (1949) of isothermal grain growth led to

$$d^2 - d_{rex}^2 = 2c_1 t\ exp\frac{-Q_{gg}}{R\ T} \tag{3.26}$$

where c_1 is a constant which includes the grain boundary specific interfacial energy, Q_{gg} is the activation energy for grain boundary migration. The other symbols have their usual meaning. The results generated by the use of equation (3.26) are in good agreement with experimental observation on growth of cells in a soap-water foam, within a partially evacuated container (Fisher and Fulllman, 1952). However, the agreement between theory and experiments in metals being annealed at high temperature has proven to be less successful (Ruibal, 1983). Sellars and Whiteman (1976)

analysing data from Miller(1951) and Kozasu et al.(1975) arrived at the following relationship for grain growth of C-Mn and Nb-steels at short times:

$$d^{10} - d_{rex}^{10} = A' \ t \ \exp \frac{-Q_{gg}}{R \ T}$$ (3.28)

where A' and Q_{gg} are constants and the other symbols have their usual meaning.

The high value of the time exponent was found to correlate the available experimental data well, as it can be seen from figure (3.12). It is however, acknowledged (Sellars,1980) about equation (3.28) that:

a- It is not considered to be satisfactory in principle.

b- It implies that grain sizes below a certain limit can not be maintained even for very short periods of time after recrystallization.

c- If $d_{rex}^{10} \ll A' \exp (-Q_{gg}/ RT)$ and if the time for full recrystallization to take place is short compared with interpass times, then the entering grain size in the next stand depends almost entirely on the temperature.

In order to overcome the inadequacies presented by equation (3.28), it has been suggested that equation (3.28) should be modified to a form in which the time exponent is much lower than 10 and that the constant A' should vary with time (Sellars,1980).

Following the argument put forward by Sellars (1980), an equation of the form

$$d^2 - d^2_{rex} = kt \qquad\qquad (3.29)$$

where

$$k = k_s \left[1 + \beta \exp\left[\frac{-t}{\tau}\right]\right] \qquad\qquad (3.30)$$

and k_s, β and τ are constant at a given annealing temperature has recently been proposed (Foster,1981). It has been used to fit a wide range of data on isothermal grain growth of C-Mn and Nb-steels with reasonable success as it can be seen from figure (3.13).

A time exponent lower than 10 has been frequently reported in the literature for austenitic stainless steels (Nörstrom,1977; Martin and Holmes,1979; Towle and Gladman,1979). Analysis of the available experimental data published on grain growth of austenitic stainless steels (Nörstrom,1977; Towle and Gladman,1979) shows that k in equation (3.29) is approximately independent of the annealing time as shown by figure(3.14). In this way, k can be written as

$$k = A'' \exp\frac{-Q'}{R\,T} \qquad\qquad (3.31)$$

where $A''=3.10 * 10^{13}$ and $Q'=320$ kJ/mol.

Theoretical predictions using equation (3.29) are in reasonable agreement with the experimental data available on AISI316 as shown in figure (3.15).

CHAPTER 4

Hot Rolling Simulation

4.1 Computer Simulation.

Hot working operations usually involve deformation of the metal to high strains at strain rates in the range of 1 to $1000s^{-1}$ at temperatures above approximately 0.6 of the absolute melting temperature. In a typical hot rolling schedule, increments of equivalent-true strain of 0.2 to 0.7 are given in each pass followed by time intervals of cooling in which the material travels from one stand to the following one. During resting periods, the material undergoes structural changes due to static recovery and recrystallization. These structural changes will determine the structure before the following pass as well as the mechanical behaviour of the material during rolling.

Simulation of complex changes which occur in a hot rolling schedule was first attempted by the use of a simple numeric algorithmic which describes the hypothetical structure evolution during an isothermal rolling schedule with passes of constant strains (ε = 0.6) given at constant time intervals between passes (Sellars,1974; Sellars and

Whiteman,1974). Although unrealistic, this approach showed
that the time for recrystallization in between passes is
critical in determining the final structure of the hot
rolled material. It also showed that when the time for
static recrystallization to take place is of the order of
the inter-pass time interval, the structures developed after
successive passes are mixed in terms of both grain size and
stored energy.

Experimental data on basic relations for structural
evolution were reviewed by Sellars and Whiteman(1976). This
review allowed a more extended version of the semi-
quantitative approach described previously to be further
developed.

The simulation of grain size evolution of a simplified
controlled rolled schedule for both mild and Nb-steels was
performed (Sellars and Whiteman,1976). The inter-pass times
and the heat loss rate were assumed constant during rolling.
Nevertheless, the authors have reported satisfactory
agreement between theoretical predictions and reported
values in the literature.

Using equations previously developed for C-Mn and
Nb-steels, Sellars and Whiteman(1979) produced a critical
analysis of their mathematical model used for computation of
microstructure evolution during hot rolling. They considered
the effects that experimental uncertainties in defining the
basic structural equations for time to 50%
recrystallization, recrystallized grain size and grain
growth may have on the predictions of the resulting

microstructure from a particular hot rolling schedule. They also examined the effects of processing variables such as pre-heat temperature, original grain size, cooling rate and amount of strain given in the last pass on the development of the microstructure. The authors showed that experimental errors in the determination of basic equations have relatively small effects in the final structural predictions. This can be explained in terms of a self-compensating effect occurring during the hot rolling schedule. Hence, if any factor led to a faster recrystallization or grain growth rate, it would lead to a larger grain size after one pass, tending therefore to delay the recrystallization after the following pass. This self-compensating effect may also explain the fact that though working with a hypothetical hot rolling schedule, the final predicted grain sizes are in reasonable agreement with observed ones in the industrial practice. Somewhat surprising, however, is the negligible effect that multiplying the recrystallization rate by a factor of 5 has on the final microstructure.

The structure model so developed has been enhanced (Leduc,1980) by merging it with a computing model for predicting temperature evolution during hot rolling (Harding,1976). At the same time, basic laboratory data on the stress-strain curves for C-Mn and Ti-bearing steels were incorporated in the model so that predictions of rolling loads and power consumption could be made. There is good agreement between predicted values of fraction

recrystallized and recrystallized grain sizes with the ones experimentally obtained from laboratory hot rolling mill samples which were either fully recrystallized or fully deformed after a certain pass had been given. When partial recrystallized data are available, the agreement seems to be less successful. This may be attributed to difficulties in measuring fraction recrystallized in transformable steels.

An alternative way of simulation (Le Bon and Saint-Martin,1977) consists of using a mathematical model which takes into account the hot strength of the material for designing an optimum schedule for a given plant installation, maximising productivity. The schedule so designed must then be checked by means of a hot torsion test to provide for structural evolution and stress-strain behaviour information. If the structure obtained is not considered to be satisfactory, alteration is made on the previous schedule and a new simulation test run is performed. The procedure is repeated until a suitable solution is found.

The Zener-Hollomon parameter is a linear function of the strain rate and an exponential function of the temperature. Clearly, although both parameters should be monitored accurately, it is of more importance to define the temperature profile during hot rolling with some precision. Cornfield and Johnson (1973), using a finite-element method for modelling, indicated the strain and strain rate variations through the roll gap which might arise as a consequence of different temperature distribution in the

rolling slab.

Modelling of the temperature distribution during hot rolling has been accomplished by numerical methods (Harding,1976; Shepard and Wright,1980; Tseng,1982) and analytical ones (Bryan and Heselton,1982). However, only few investigators have extended their findings about temperature distribution to attempt to analyse its effects on the microstructural evolution of the material being rolled (Sheppard and Wright,1980).

4.2 Mechanical Tests.

4.2.1 Tensile Test.

The necessary equipment to perform a tensile test is usually widely available. The testing technique is well known and there are no frictional problems involved. The material, however, undergoes uniform deformation for an interval of strain which is approximately 0.2 to 0.3 for steels (Wallquist,1962). After that, necking occurs and the strain and the strain rate become localized to a small region of the sample. Rossard has constructed a special tensile machine in which the rate of the pulling is adjusted with respect to the necking behaviour of the material undergoing test by using a logarithmic cam. In this way, the

strain rate in the necking region remains constant (Rossard and Blain,1959). On this machine, strains up to 2.0 have been reported as having being attained, and stress-strain curves are similar to the ones obtained in torsion. However, when necking appears, the state of stresses is no longer uniaxial.

In general, the decreasing strain rates, as the sample is deformed before necking (if the cross head speed is kept constant during testing), and the necking occurring at relatively small strains rule out tensile testing for hot rolling simulation purposes (McQueen and Jonas,1971).

Tensile tests, however, have been used in the determination of the recrystallization kinetics (Liljestrand,1972; Sveson,1977; Wray,1977; Nörstrom,1977) and restoration studies using double deformation technique (Cordea and Hook,1970).

4.2.2 Axisymmetric Compression.

In this kind of test, a cylindrical sample is compressed in between platens. The strain rate during testing increases with the deformation imparted to the sample, if the cross head speed is kept constant.

Axisymmetric compression tests have been commonly used for measuring ductility of materials. However, there have not been reports of axisymmetric compression being used for simulation of hot rolling. This has been largely attributed to the fact that bulging occurs in the testing samples at

strains of ~0.7 (McQueen and Jonas,1971). Bulging is caused by friction between the sample surfaces and the tools. A number of lubricants have been used for different materials and test temperatures (Sellars and Tegart,1972).

When the sample bulges, the strain is no longer homogeneously distributed. Recently, axisymmetric compression has been used to simulate forging conditions which induce shear bands inside the material being deformed (Semiatin and Lahoti,1983a). Axisymmetric compression tests have also been reported to be used in the study of kinetics of recrystallization (Petkovic et al.,1977; Ahlblom,1977a, 77b; Guimaraes and Jonas,1981) and precipitation in Nb-steels and low carbon steels (Weiss and Jonas,1979)

4.2.3 Torsion Test.

Torsion tests have been used for determining ductility of metals and to simulate hot rolling. Although the sample is theoretically deformed by shearing, deformation does not proceed during testing without a change in the length of the sample (Moore,1968). It has been previously thought that large strains (up to equivalent strains of 20) could be imparted to the sample without having plastic instability (McQueen,1968). Recently, it has been shown that unstable flow may be initiated at a much lower strain causing a drop in the stress-strain, curve since flow localisation enables the deformation to continue at torques less than would be required for uniform

deformation (Semiatin and Lahoti,1981b).

During hot torsion, the strain and strain rate vary proportionally to the radius of the sample. As shear stress is a function of the strain and strain rate, there will be a stress gradient through the sample. Gradients in stress, strain and strain rate give rise to the existence of microstructural gradients through the gauge length. Barraclough et al.(1973) proposed the use of an 'effective radius' for the calculation of the stress, strain and strain rate from the torque-twist curve. It has also been shown that it could be advantageous to measure any metallographic feature over the 'effective radius' since they can be directly related to the flow stress calculated over that radius.

Without doubt, the biggest advantage of the torsion test is that it may be carried out without any frictional problems present in the other tests. Therefore, it has been widely used for the determination of mechanical properties of metals (McQueen et al.,1977). The torsion test has also been reported to be used for simulation of hot working operations such as rolling (Cuddy,1977; Gittins et al.,1977; Knudsen et al.,1980; Radu et al. 1979a; 1979b; 1981) and hot extrusion (Soòs et al., 1981a; 1981b). Most of the simulation work carried out was performed isothermally. In contrast, a few non-isothermal simulations have been reported in the literature (Rossard et al.,1962; Vaughan,1968; Farag et al.,1968, Migaud,1980). This may be because of the experimental difficulties involved in

relating stress and microstructural observations to a meaningful testing temperature.

4.2.4 Plane Strain Compression Test.

In this test a strip is compressed in between a pair of tools. The Ford test (Ford,1948) was based on this arrangement. It was later established (Watts and Ford,1952-53,1955) that the testing technique for reproducible and accurate results should include:

a- Incremental loading in the same indentation, with lubrication and measurement of strain at each increment. Increments should be ~2% reduction each.

b- The ratio w/h should be maintained between 2 and 4, dies being changed during test.

c- The width of the strips should be at least 6 times the die breath, preferabily more.

Conditions a and b cannot practically be fulfilled during hot plane strain compression tests. Therefore, with a fixed tool width, the test becomes limited to maximum strains of ~2.5, since above this limit frictional effects become large, even with efficient lubrication (Sellars,1981). Therefore, the major drawback of plane strain compression is friction. Its effects increase during testing, since the ratio width:thickness of the sample increases. In order to

calculate the stresses from the pressure versus displacement curve, allowances for friction must be made. If however, the friction coefficient is kept low (sliding friction), the pressure exerted by the tools can simply be calculated as (Hill,1950)

$$\bar{p} = \sigma ' \left[1 + \mu \frac{w}{2h} \right] \qquad (4.1)$$

where μ is the friction coefficient and σ' is the stress in plane strain.

For sticking or partially sticking friction the equations required are somewhat more complex. These have been previously reviewed (Sellars et al.,1976) and are reproduced in section(5.37).

Plane stress, strain and strain rate are converted to equivalent equiaxial ones using a modified Von Mises criterion (Sellars et al.,1976) which takes into account the departure from plane strain conditions observed during test.

The major advantage in the use of the plane strain compression test is that it imparts on the sample the same mode of deformation as in rolling. The test has been frequently used for measuring mechanical properties and for kinetics studies (Bailey and Singer,1963-64; Evans and Dunstan,1971; Leduc and Sellars,1981) and it has been increasingly used for hot rolling simulation (Leduc,1980; Foster,1981; Puchi,1983)

4.2.5 Hot Rolling.

As a method of simulation, laboratory scale hot rolling is the only type of test capable of imparting to the sample the same deformation mode as the one found in industrial processes. However, there are several practical difficulties which may have caused the test to be rarely used for simulation purposes:

a- Control of the test parameters and reproducibility are usually more difficult than for any other of the previously mentioned tests.

b- Due to the size of the specimen, simulation of realistic cooling rates are made very difficult.

c- Friction conditions between sample and working rolls are different from the ones found in mill operations.

Therefore, the biggest problem with laboratory scale hot rolling is the difficulty of extending the simulation findings to production mills. However, the test has been used for obtaining basic information on the material behaviour in hot rolling (Sheppard and Wright,1980; Brown and DeArdo,1981; Zaidi and Sheppard,1982). There have also been reports of laboratory hot rolling being used for kinetics of static recrystallization studies (Towle and Gladman,1979).

4.3 Concluding Remarks.

It can be seen, in summary, that most laboratory tests do not closely simulate the conditions of a rolling pass. Except for laboratory scale hot rolling, all the other tests present flow instability at a certain stage during the deformation process. Non-homogeneous strain distribution over the gauge section may induce structural gradients which may, in turn, be a great problem during simulation.

The main variables to be closely monitored during hot rolling simulation are the initial grain size, temperature evolution, strain, strain rate and time. As simulation is not a reproduction but a modelling of real processes, it is necessary to adopt some criteria of equivalence between laboratory and industrial process in order to obtain a correct evaluation of the hot workability and structural paramenters (Radu et al.,1981). It is important to ensure that the appropriate microstructure for the pass of interest is present in the specimen (Sellars,1981). This equivalence in microstructure means having a convenient reproduction of the cooling rate and timing during the rest periods. During the deformation, it has been proposed that an equivalence criterion may be (Radu et al.,1979b)

$$\epsilon^p = \epsilon^t \qquad\qquad (4.2)$$

$$L^p = L^t \qquad\qquad (4.3)$$

where the superscripts p and t mean process and test respectively.

$$L = \int_0^{\varepsilon_f} \sigma d\varepsilon = \bar{\sigma} \, \varepsilon_f \qquad (4.4)$$

and since it has been found (Barraclough and Sellars,1974) that during high temperature deformation under constant conditions, flow stress at all strains is determined by the temperature-compensated strain rate, and furthermore, that an equation of state successfully predicts stress-strain curves, it can be said that the criterion (4.4) may be rewritten as

$$z^p = z^t \qquad (4.5)$$

If the average test temperature coincides with the average process temperature, then (4.5) transforms into

$$\dot{\varepsilon}^p_{av} = \dot{\varepsilon}^t_{av} \qquad (4.6)$$

which is largely used as a simulation criterion in the literature instead of (4.5).

CHAPTER 5

Experimental Procedure

5.1 Introduction.

In the previous chapters, it has been shown that:

a- It is not completely clear how the rate of static recrystallization can be described in terms of the processing and material parameters. This is particularly true for the t_{50} dependence on the sample original grain size. It has also been shown that the strain distribution may have an important part to play in the softening behaviour of the material.

b- It has been revealed that little work has been done on simulation of hot rolling when the time for static recrystallization of the sample being deformed is of the order of the inter-pass period.

The aim of the present research is to simulate hot rolling of austenitic stainless steel AISI316L. This material has been chosen for two main reasons:

a- The material does not transform when cooled to room temperature, therefore metallographic observations are relatively easy.

b- The material may present stages during the deformation process in which a partially recrystallized structure can occur in between passes.

The present research programme is schematically depicted in figure(5.1). This involved:

a- Hot rolling of as received material for generation of basic information on hot rolling of AISI316L and for the preparation of samples for plane strain compression samples.

b- Assessing the basic data needed for numerical simulation of microstructural evolution. This involved studies on:

1- Kinetics of static recrystallization

2- Recrystallized grain size

3- Grain growth

4- Strength during hot rolling

c- Simulation of hot rolling by using a computer model put forward by Leduc(1980).

d- Simulation of hot rolling by direct use of non-isothermal plane strain compression tests.

5.2 Hot Rolling.

5.2.1 Specimen Preparation.

As mentioned earlier, the material used in the present investigation is AISI316L, whose chemical composition can be seen from table (5.1). The original material was received as bars of approximately 6m in length with nominal cross section of about 51x25 mm (2"x1").

Samples for hot rolling schedules were cut from the bars into slabs with nominal length 100 mm.
Whenever temperature was to be measured, a single Pyrotenax thermocouple (nominal diameter 1.5 mm) inserted in a drilled hole was always positioned at mid-length, mid-width and mid-thickness of the sample as can be seen in figure(5.2).
The measurements of the initial thickness and width were taken at the region where the thermocouple was implanted. At least three different measurements for each dimension were carried out.

5.2.2 Hot Rolling Tests.

The hot rolling experiments were carried out on a fully instrumented 50 tonnes, two-high laboratory rolling mill. The mill has two rolls of 139mm diameter each.
The rolling loads are measured by two load cells of 25 tonnes capacity each, positioned between the end of the

screw down and the chocks of the top roll. The load cells'
output is fed via an amplifier into a U.V. recorder
(SE3006). The deflection obtained is converted into loads by
using calibration charts.

The roll speed is measured by means of an aluminium disc
with eight equally spaced holes drilled on a circle of 50 mm
diameter. The disc is fitted at the undriven end of the
bottom roll. The disc is mounted so that it is positioned in
between a lamp and a photocell.

The output from the photocell is fed into the U.V. recorder,
which registers peaks of maximum output intensity. From the
distance between the peaks on the U.V. recorder chart, the
peripheral roll speed, V, can be calculated as

$$V = \frac{S\pi R}{4 d} = n\ 2\pi R \qquad (mm/s) \qquad\qquad (5.1)$$

where S is the chart speed in mm/s, R , the roll radius, d
is the separation distance of the peaks and n is the number
of revolutions per second of the bottom roll.

During a test run, the thermocouple output in fed into a
Telsec recorder of 0.5s response time. The Telsec recorder
is always calibrated before a run by using a potentiometer.
Temperature measurements during hot rolling experiments were
carried out without using an ice-water cold junction.

Samples were re-heated in air in a muffle furnace for ~15min
prior to each run. The passes were given at constant time
intervals of 15s and the samples were water quenched
immediately after the last pass. The time elapsed between
finishing the last pass and the quenching operation was of

the order of 2s.

5.2.2.1 Kinetics of Static Recrystallization.

In order to study the kinetics of static recrystallization from hot rolled samples, four slabs were cut with initial dimensions 100x25.5x25.5 mm. The specimens were cut to half their usual width so that the effect of rolling reduction on the recrystallization rate could be studied over a wider range of reductions. The single pass schedules consisted of nominal reductions of 15,25,30 and 35%.

Each slab was heat treated at 1180 C for ~50 minutes, which produced an original grain size of 191±15 μm. After heat treatment, the samples were quenched, shot blasted for scale removal, drilled and a thermocouple was inserted in each slab.

Just before hot rolling, the slabs were pre-heated at ~980 C for a period of time sufficient for temperature homogenization (normally ~15 minutes). A pass was then given when the temperature measured at the centre of the specimen was ~950 C. The hot rolled slabs were immediately quenched and sectioned as shown in figure (5.3a).

The samples were annealed at 950 C for several periods of time. For annealing times less than 600s, a thermocouple was implanted in the sample and an e.m.f. versus time curve was

registered on a Telsec recorder. An annealing equivalent-time for recrystallization at 950 C was calculated using (Whittaker,1973)

$$W = \sum_{0}^{t} \left[\exp- \left[\frac{Q_{rex}}{R\ T} \right] \right] \delta t \qquad (5.2)$$

and assuming that $Q_{rex} = 525$ kJ/mol (Towle and Gladman,1979).

5.2.2.2 Strength and Simulation of Hot Rolling.

The standard hot rolling schedule consisted of three passes of 25% reduction each with interval times of 15s. All the slabs were pre-heated so that the original grain size was either 100μm or 200μm.

Before hot rolling, the slabs were held in air for a period of time necessary to give a desired measured centre temperature at the entry to the first pass. If the pre-heat temperature and the measured centre temperature at the entry to the first pass differed by more than ~100C, a second furnace set at a lower temperature was used. In this manner, temperature gradients due to prolonged air cooling periods were eliminated.

Loads and temperature were measured during test and processed later as detailed in section (5.2.5). Samples were also rolled by one and two passes and quenched after 15s air cooling in order to retain the microstructure for fraction recrystallized and average recrystallizing grain

size measurements. This experimental rolling procedure is shown schematically in figure (5.4).

Some double pass schedules were arranged as to have the first pass given at a measured centre temperature ranging from 1200 to 1030 C. An air cooling period was allowed in between passes 1 and 2 in order to have the temperature entering the second pass at ~950C. In this way, the effect of the fraction recrystallized entering pass no. 2 on the rolling load could be studied.

5.2.3 Optical Metallography.

Optical metallography was carried out in a section a distance 'y' from the side surface, where

$$y = 0.211 \ w \qquad\qquad (5.3)$$

This position, shown in figure(5.2), is chosen because it has a temperature history close to the average for the cross section (Leduc,1980). In equation(5.3), "w" is the average final width of the slab. The samples were mechanically polished through a series of successively finer silicon carbide papers, followed by a 6 μm diamond polish. The finishing was given with a 1 μm diamond polish. The specimens were then electrolytically etched using concentrated nitric acid at 4V for ~30s at room temperature (Cole,1979).

5.2.4 Quantitative Metallography.

The assessment of the fraction recrystallized was carried out using a semi-automatic Swift Point Counter, mostly at a magnification of x100 and using a step increment length of 0.05mm. The number of points counted were such as to give a maximum relative error ~0.05 when calculated as (Sellars,1981c; Gladman and Woodhead,1960)

$$\frac{\sigma_{vf}}{V_f} = \sqrt{\frac{1 - V_f}{N_f}} \tag{5.4}$$

where N_f is the number of points counted in the recrystallized region, V_f is the volume fraction recrystallized and σ_{vf} is its standard error. For a typical run, if the total number of points counted is 600 and the fraction recrystallized, V_f, is 0.25 then, the relative error is 0.035. The standard error for this fraction recrystallized is

$$\sigma = 0.035 \times 0.25 = 8.75 \ 10^{-3}$$

thence the 95% confidence limit is about 2×10^{-2}, therefore,

$$V_f = 25 \pm 2 \ \%$$

is a value to be tabulated.

The measurements of fraction recrystallized were carried out mostly on lines a, b, and c, as shown in figure (5.3-b). In samples used for kinetics of static recrystallization studies, measurements were taken on lines a', b' and c' as detailed in figure (5.3-a).

The grain size was measured using a Beck microscope equipped with a micrometric travelling stage. The mean-linear intercept method was used. Materials which were fully recrystallized or annealed were counted over lines parallel to the rolling direction. A sample more than 80% recrystallized was considered 'fully recrystallized' since it has been shown (Leduc,1980) that the average recrystallizing grain size is very similar to the actual recrystallized grain size (within~10%) if V_f is greater than approximately 60%, as shown in figure(5.5). Partially recrystallized structures were measured on lines parallel and perpendicular to the rolling direction. A geometrical average was taken as the mean linear intercept corresponding to the mean recrystallizing grain size.

It has been shown (Silva,1966), from experimental observations, that

$$\frac{\sigma (\bar{N}_1)}{\bar{N}_1} = \frac{k}{\sqrt{N}} \tag{5.5}$$

where k is a constant equal to 0.65 ± 0.02 and N is the number of grains counted. Accordingly, if N=200, the relative error is 0.045(4.5%). The standard error for a mean linear intercept of 100 μm is thus 4.5 μm and the 95% confidence limit is 9.0 μm. Therefore, the mean linear intercept is written,

$$\bar{N}_1 = (100 \pm 9.0)\ \mu m.$$

For partially recrystallized samples, the 95% confidence limit is given by:

$$\Delta = N_1 \left[RE_t + RE_1 \right] \tag{5.6}$$

where RE is the relative error and the subscripts t and l stand for transversal and longitudinal directions.

5.2.5 Data Handling.

Measurements of the final width and thickness were made at the cross section were the thermocouple had been imbedded. At least 3 measurements of each feature were taken. The mean final width was calculated by using (El-Kalay and Sparling,1968)

$$W_2 = w_r - \frac{1}{3} (w_r - w_b) \tag{5.6}$$

where w_r is the mean width over the crown and w_b is the mean width measured at the top and bottom surfaces as shown in figure (5.6).

Predictions of the intermediate values of w_2 during a multi-pass schedule were performed as follows (see section (8.4)):

$$w_{(i+1)} = w_i \exp S\varepsilon' \tag{5.7}$$

where S is the lateral spread occurring during a given pass and is defined as :

$$S = 0.054 \ \exp\left[1.336 \ \frac{h_i}{\sqrt{R \ \Delta h}}\right] \qquad \text{if} \quad w_i < 40\text{mm} \qquad (5.8)$$

and

$$S = 0.061 \ \left[\frac{h_i}{w_i}\right]^{1.3} \ \exp\left[1.265\frac{h_i}{\sqrt{R\Delta h}}\right] \qquad \text{if} \quad w_i > 40\text{mm} \qquad (5.9)$$

Formulae (5.8) and (5.9) are similar to the ones proposed by Beese(1972).

The equivalent true strain and strain rate are given by

$$\varepsilon = \frac{2}{\sqrt{3}} \ \ln\left[\frac{h_i}{h_{i+1}}\right] \qquad (5.10)$$

and

$$\dot{\varepsilon} = V\left[\frac{1}{\sqrt{R\Delta h}}\right] \varepsilon' \qquad (5.11)$$

respectively (Larke,1963).

The peripheral roll speed , V, was kept constant and equal to 203 mm/s during the rolling experiments. The mean strain yield stress, k, was calculated by using Sims' theory(1954);

$$P = k \ b \ \sqrt{R\Delta h} \ Q \qquad (5.12)$$

where

$$Q = \frac{\pi}{2} \sqrt{\frac{1-r}{r}} \tan^{-1} \sqrt{\frac{r}{1-r}} - \frac{\pi}{4} - \sqrt{\frac{1-r}{r}} \sqrt{\frac{R'}{h_2}} \ln \frac{h_n}{h_2} \qquad (5.13)$$

$$+ \frac{1}{2} \sqrt{\frac{1-r}{r}} \sqrt{\frac{R'}{h_2}} \ln \frac{1}{1-r}$$

and

r = fractional reduction

b = stock width

P = rolling load

R = roll radius

R'= elastically deformed roll radius

Δh= draft

k = mean plane strain yield stress

h_n= thickness at the neutral plane

Additionally,

$$h_n = h_2 \left[\frac{R'}{h_2} \alpha^2 + 1 \right] \qquad (5.14)$$

where

$$\alpha = \tan \left[\frac{\pi}{8} \left(\frac{h_2}{R'} \right)^{1/2} \ln(1-R) + \frac{1}{2} \tan^{-1} \left[\sqrt{\frac{r}{1-r}} \right] \right] \sqrt{\left(\frac{h_2}{R'} \right)} \qquad (5.15)$$

During calculations of the rolling load, an average width was used. Therefore,

$$b = \frac{w_i + w_{i+1}}{2} \qquad (5.16)$$

It has also been assumed that

$$R' \simeq R \qquad\qquad\qquad (5.17)$$

since the corrections introduced for roll flattening are small even at relatively high stress levels (Harding,1976) and are often omitted. The mean plane strain yield stress was converted into the mean equivalent flow stress, $\bar{\sigma}$, by using the Von Mises criterion

$$\bar{\sigma} = \frac{\sqrt{3}}{2}k = \frac{k}{1.155} \qquad\qquad (5.18)$$

If the sample has lateral spread during hot rolling, the flow stress will lie in between the one for axisymmetric compression and the one for plane strain. The corrected mean plane strain stress can then be calculated from (Leduc,1980)

$$k = \bar{\sigma}\sqrt{1 + S' + S'^2} \qquad\qquad (5.19)$$

S' is given by

$$S' = \frac{\ln\left[\frac{w_{i+1}}{w_i}\right]}{\ln\left[\frac{h_i}{h_{i+1}}\right]} = \ln\frac{\left[\frac{w_{i+1}}{w_i}\right]}{\varepsilon'} \qquad\qquad (5.20)$$

An annotated version of the computer programme used for the calculation of the mean plane strain yield stress is presented in appendix (5.1).

5.3 Plane Strain Compression Tests.

5.3.1 Specimen Preparation.

As mentioned previously, slabs were hot rolled for the purpose of preparing plane strain compression samples. The hot rolled slabs finished with ~10.8mm thickness after the usual 3x25% reduction schedule. The plates were cut into lengths of ~65mm and heat treated to give a desired grain size, nominally :

1/2 h at 1200 C	~100 μm
3/4 h at 1300 C	~260 μm

After heat treatment, the samples were machined using a vertical milling machine with carbide tip cutters. The feeding speed (~3"/min or 1.28 mm/s) was such as to produce machine grooves in the surface of the sample adequate to retain lubricant during testing. The geometry of the specimen is shown in figure(5.7).

The lubricant used during isothermal tests was Dag2626 provided by Acheson Colloids Co.. The glass was painted on the sample surface. The best lubrication effect seems to be achieved if two thin layers of lubricant are applied. Excess lubrication shortens the tools' testing life. Several combinations of glass lubricant have been tried for lubrication of non-isothermal tests (F1237 + Dag2626, F1237 + Dag2626 + Dag2251, F1237 + Dag2651, F1237 + Dag57, Dag11). However, Dag2626 still seemed to give the

best lubrication characteristics and oxidation protection.

5.3.2 Stress-Strain Curves.

The stress-strain curves for the AISI316L were obtained from plane strain and axisymmetric compression tests. The samples were tested using a Servotest computer controlled, servo-hydraulically driven testing machine. The machine and the ancillary equipment have been comprehensively described elsewhere (Sellars et al.,1976; Beynon,1979; Foster,1981).

The tests were performed at 950, 1025 and 1100 C with strain rates ranging from 0.05 to $5s^{-1}$ and original grain sizes of 100 and 260 μm. The pre-heating procedure was made up of a soaking period of ~15 minutes at temperatures of 980,1150 and 1130 C in the re-heating furnace followed by 180s soaking inside the testing furnace before each test, in order to eliminate thermal gradients introduced during transfer between furnaces. The initial and final width and thickness were corrected for thermal expansion. The dependence of the average expansion coefficient from room to test temperature on testing temperature is shown in figure (5.8). Temperature measurements were carried out by using a dummy sample with an embedded thermocouple at the beginning and at the end of each testing day. Data for load, displacement and millivolts output were recorded in paper tape and converted into stress-strain curves corrected for origin, friction and spread (Foster,1981).

5.3.3 Kinetics of Static Recrystallization.

Tests were carried out at 950,1025 and 1100 C at strain rates of $5s^{-1}$ and original grain sizes of 100 µm and 260 µm. Samples were subjected to a prior strain which included values smaller and greater than the strain to peak stress. Holding times varied from nil to ~900s. Samples with annealing times longer than 900s were quenched, cut and heat treated with an embedded thermocouple. The equivalent time at a given test temperature was calculated using equation (5.2).

The samples were normally sectioned at the mid-width position as shown in figure (5.9a). Metallographic preparation was similar to the one carried out for hot rolled samples. Measurements of fraction recrystallized were obtained by point counting at lines a,b and c as shown in figure (5.9b). Care was taken to avoid the transitional region between the gauge section and specimen shoulder. The number of points counted were such as to give a maximum relative error of ~0.05. Typically, at least 600 counts were performed per sample.

5.3.4 Recrystallized Grain Size.

Samples tested for static recrystallization studies were used to obtain information on the average recrystallized grain size. Specimens deformed at temperature less than 1100 C were quenched and annealed at the test temperature for periods of time sufficient to give fully

recrystallized structure. Grain growth at 950 and 1025 C was expected to be negligible. This was later investigated. Experiments conducted at high temperature (1100 C) had their average recrystallized grain size counted from specimens with no less than 80% recrystallized fraction. Average recrystallized grain size was measured over an area of ~4x2 mm in the centre of the gauge area, as shown in figure (5.10). Whenever possible, a number of grains was counted so as to give a maximum relative error ~0.05.

5.3.5 Grain Growth Tests.

A number of grain growth measurements were carried out at 1025 and 1100 C after performing tests with a constant strain rate of $5s^{-1}$ on samples with original grain size 100 μm. The samples were pre-heated at ~1050 and 1130 C for a period of time of ~10 minutes. They were then transferred to the testing furnace where they were left for at least 180s before testing. A single deformation was given and the sample was held inside the testing furnace for a pre-selected period of time before being quenched. If the time exponent in the Avrami equation, k, is assumed to be ~1, then the time for 95% recrystallization can be written as

$$t_{95} = 4.32 \ t_{50} \hspace{4cm} (5.22)$$

The time for grain growth was then calculated as

$$t_{gg} = t_a + t_b - t_{9s} \qquad\qquad (5.23)$$

where

t_a = nominal annealing time, seconds

t_b = time, in seconds, spent in transporting the sample from test to quench position. It was assumed to be ~2.0s (Foster,1981).

After quenching samples were sectioned, polished and etched. The final grain size was measured as described in section (5.3.4).

5.3.6 Interrupted Deformation Tests.

Double deformation tests were carried out to yield information on the kinetics of static recrystallization. It was thought that the recrystallization process would be affected by the concentration of strain in the shear bands. The samples were tested at 950 and 1025 C at $5s^{-1}$ with an original grain size of 100 μm. The strain given in the first stroke varied from 0.33 to 1.6. The restoration index was calculated as(Barraclough and Sellars,1979)

$$R = \frac{\sigma_2 - \sigma_3}{\sigma_2 - \sigma_1} \qquad\qquad (5.24)$$

where

σ_1 = yield stress in the first deformation(2% offset)

σ_2 = maximum stress in the first deformation

σ_3 = yield stress in the second deformation (2% offset)

5.3.7 Data Handling.

After a given test, the raw data (load, displacement, millivolts) were stored on paper tape. A computer programme has been developed (Foster,1981) to treat the raw data converting them into stress, strain, strain rate and temperature. A detailed description of the programme is provided elsewhere (Foster,1981). In the present section only a brief outline of the calculations and the corrections involved are given.

The millivolt readings from the embedded thermocouple were converted into temperature by(Sellars et al.,1976)

$$T = -5.781 + 26.28V - 0.125V^2 + 0.0019V^3 \qquad (5.26)$$

where T is the temperature in degrees Celsius and V the millivoltage output. Expression (5.26) is only valid for chromel-alumel thermocouples.

The pressure exerted by the tools is given by

$$\bar{p} = \frac{L}{A} \qquad (5.27)$$

where L is the load and A the instantaneous contact area between tool and sample. In reality, some lateral spread occurs during testing. Therefore, A is dependent on the instantaneous thickness of the sample. The width b for a given value of specimen thickness h can be calculated from (Sellars et.al.,1976)

$$\frac{b}{b_o} = 1 + c - c \sqrt{\frac{h}{h_o}} \qquad (5.28)$$

where b_o and h_o are the initial breath and thickness respectively. The coefficient of spread, c, is experimentally determined. The value of c is 0.33 for AISI316L as it can be seen from figure (5.11).

If only sliding friction is present during test, the pressure exerted by the tools may be modified by using

$$\frac{\overline{p}}{2 k} = \frac{h}{\mu w} \exp\left[\frac{\mu}{h} w - 1\right] \qquad (5.29)$$

where μ is the friction coefficient, w the tool width and k the yield stress in pure shear. If sticking friction is present over the whole of the tool surface, then

$$\frac{\overline{p}}{2 k} = 1 + \frac{w}{4 h} \qquad (5.30)$$

However, it may happen that both sliding and sticking friction occur during test. In that case the pressure may be corrected by using

$$\frac{\overline{p}}{2 k} = \left[\frac{h}{\mu w}\right] \left[\frac{1}{2 \mu} - 1\right] \left[\frac{\frac{w}{2} - z_o}{\mu w}\right] + \frac{\left[\frac{w}{2} - z_o\right]^2}{h w} \qquad (5.31)$$

where

$$Z_o = \frac{h}{2\mu} \left[\ln \left[\frac{1}{2\mu} \right] \right] \qquad (5.32)$$

If $Z_o > w/2$, only sliding friction will exist during the test. On the other hand, if $Z_o < w/2$ there will be sticking friction in a central band on the tool surface $(w-2Z_o)$ wide and sliding friction outside this band.

The equivalent uniaxial yield stress is calculated using Von Mises criterion modified to take into account the spread (Sellars et al, 1976)

$$Y = \frac{2\ k}{f} \qquad (5.33)$$

where

$$f = \frac{(1.155\ (b - w) + w)}{b} \qquad (5.34)$$

Accordingly, the equivalent true strain is given by

$$\varepsilon = f\ \ln \left[\frac{h_o}{h} \right] \qquad (5.35)$$

The strain rate is computed as

$$\dot{\varepsilon} = \frac{\varepsilon}{t} \qquad (5.36)$$

The interval of strain over which the strain rate is evaluated is constant and equal to 0.05. The corresponding time interval t is determined since the clock frequency is a known parameter (Foster, 1981).

Further corrections proposed for the pressure vs. strain curve are the ones for origin, temperature and original

geometry.

The origin correction used in the present work is the one suggested by Colàs(1983). In this correction procedure, a straight line is interpolated in the initial portion of the load vs. displacement curve as shown in figure (5.12). This straight line is interrupted by a vertical line at

$$D = R_1 - R_2 \qquad\qquad (5.37)$$

where R_1 is the maximum ram displacement read from the machine command input signal whereas R_2 is the measured displacement $(R_2 = h_o - h_f)$. The origin correction is carried out by subtracting from the displacement value for each point the value of displacement at the same load from either one of the appropriate straight line (Colas,1983).

If the experimental stress-strain data, can be correlated by an exponential type of law, then it has been shown (Sellars et al.,1976) that

$$\sigma_2 = \sigma_1 + \frac{Q}{\beta R} \left[\frac{1}{T_2} - \frac{1}{T_1} \right] \qquad\qquad (5.38)$$

where

σ_1 = the stress at a given strain to be corrected for temperature.

T_1 = nominal testing temperature,K

T_2 = average specimen temperature in the deformation zone, K

Q = activation energy for deformation.

R = universal gas constant

β = a constant

Thus, at a given constant strain rate, a stress corrected for the effects of temperature rise due to deformation can be calculated.

Stress-strain curves corrected for origin, spread, friction and temperature may still present discrepancies caused by differences in the initial testing geometry (Beynon,1979; Sellars et.al.,1976). The geometry correction incorporated in the stress-strain curve correction programme uses the following empirical equation (Beynon,1979)

$$\sigma = \sigma_o + (\sigma_i - \sigma_o) \left[1 - \frac{B \, h_o}{\varepsilon \, W} \right] \qquad (5.39)$$

where

σ_o = yield stress

σ_i = non-corrected stress at a given strain

σ = corrected stress

B = a constant

The other symbols have their usual meaning.

5.4 Non-isothermal Plane Strain Compression Tests.

5.4.1 Sample Preparation.

Sample preparation was essentially similar to the one carried out for isothermal specimens. The only difference being that a thermocouple was inserted in the mid-thickness, mid-width, mid-length position to keep a record of the specimen centre temperature. Specimens were pre-heat treated in order to give a fine structure grain size($100\mu m$) and a coarse one ($200\ \mu m$).

5.4.2 Testing Procedure.

The samples were pre-heated at temperatures ranging from 1100 to 1200 C. The soaking period was kept to a minimum necessary to give a stable homogeneous starting temperature. Typically the pre-heat time was ~8 minutes. The samples were then air cooled for 5s prior to entering the testing furnace. Times shorter than 5s were found difficult to reproduce. Unfortunately, the sample could cool down as much as 30 C during the air cooling period. Certain experiments required that the starting testing temperature would be as high as 1200 C. In these cases times shorter than 5s were attempted.

The testing furnace was kept at ~230 C. In this

way, the sample furnace cooling curve was very similar to the one obtained for the hot rolling slab cooling in air (see figure (5.13)). The average tool temperature was calculated from a diagram provided by Foster(1981) and reproduced in figure (5.14). A typical testing schedule consisted of 3x25% reduction passes with 15s furnace cooling periods between passes.

Metallographic observations were carried out on a section 0.211w of the width away from the undeformed position of the side surface. The specimens were polished, etched and point counted in the standard way.

5.5 Axisymmetric Compression Tests.

5.5.1 Specimen Preparation.

Specimens used for axisymmetric compression tests were small cylinders 7mm diameter and 10mm long. A hot rolling slab was cut into small specimens with cross section ~12.5x12.5 mm. They were machined into a bar ~7mm diameter with a turned surface finishing quality. The bar was then cut into small cylindrical specimens ~10mm height. Each of them had grooves ~0.2mm deep machined at both ends for lubrication purposes.

5.5.2 Testing Procedure.

Samples were pre-heated at ~1170 C for 1/2 hour before each test. In this manner, an original microstructure of ~100 μm grain size was generated. The transport operation between soaking furnace and testing furnace took place with a handling device schematically depicted in figure (5.15). All specimens were kept inside the testing furnace for 180s to have their temperature homogenized.

Specimens for stress-strain curves were tested at 950, 1025 and 1100 C, mostly at $5s^{-1}$.

Static recrystallization studies were conducted on samples tested at 1025 C and $5s^{-1}$ with various values of pre-strain. The specimens were quenched after testing and sectioned through a diametral plane. Quantitative metallography was carried out over the whole of the gauge section on axial lines ~1.5mm apart.

5.5.3 Data Handling.

Raw data were manipulated in much the same way as for plane strain compression samples. The assumption was made that no change in volume occurred during plastic deformation. Therefore, "A" in equation (5.27) is given by

$$A = \frac{A_o \, h_o}{h} \qquad (5.40)$$

A_o is the initial sample-tool contact area. The initial thickness is h_o and the instantaneous one is h. The pressure was corrected for friction as reported by Sellars et

al.(1976) from whom the basic formulae is reproduced below.
For sliding friction, i.e., when

$$\frac{d}{h} = \frac{1}{\mu} \ln\frac{0.577}{\mu} \tag{5.41}$$

$$\frac{\bar{p}}{Y} = \frac{2}{c^2} \left[\exp(c) - c - 1\right] \tag{5.42}$$

where

$$c = \frac{\mu \; d}{h} \tag{5.43}$$

For fully sticking friction, i.e., when $\mu > 0.577$,

$$\frac{\bar{p}}{Y} = 1 + 0.577\frac{d}{3 \; h} \tag{5.44}$$

Finally, for partially sticking friction, i.e., when

$$\frac{d}{h} > \frac{1}{\mu} \ln \frac{0.577}{\mu} \quad \text{but} \quad < 0.577$$

$$\frac{\bar{p}}{Y} = \frac{2}{c^2} \left[(D + 1) \; \exp(c - D) - c - 1 \right] \tag{5.45}$$

$$+ \frac{D^2}{c^2}\left[\frac{0.577}{\mu} + 0.577 \; \frac{dc}{3 \; h}\right]$$

where

$$D = \mu \; \frac{dc}{h} \tag{5.46}$$

and

$$dc = d - \frac{h}{\mu} \ln \frac{0.577}{\mu} \tag{5.47}$$

Y is the uniaxial tensile stress; d is the sample diameter
and other symbols have their usual meaning. For small values
of c, equation (5.42) is re-written as

$$\frac{\bar{p}}{Y} = 1 + \frac{\mu}{3}\frac{d}{h} \tag{5.48}$$

Annotated print-out listings of the programmes used for calculation of stress-strain curves are presented in appendix (5.2).

5.6 Simulation of Hot Rolling.

Simulation of hot rolling of AISI316L steel was carried out in two different ways:

a- by using a computer programme developed by Leduc(1980).

b- by direct use of plane strain compression tests.

Leduc's programme uses equations derived for isothermal microstructural changes in the material. These equations are combined with a two dimensional heat flow model and Sims' theory in order to predict fraction recrystallized, average grain size, temperatures and loads during hot rolling. The programme was originally conceived for C-Mn and Ti-bearing steels. Suitable modifications had to be introduced to the original programme as a result of using AISI316L. The alterations introduced to the programme are treated in detail in appendix (8.2).

In order to simulate hot rolling by direct use of plane strain compression tests, it was important to be able to calculate the average specimen temperature at any time during testing. A programme written by Foster(1981) was used

for this purpose. The only alterations made to the original version of the programme were related to the equations which were used for the description of heat transfer in AISI316L austenitic stainless steel. Changes carried out in the programme are detailed in appendix (8.2) together with Leduc's programme.

CHAPTER 6

Basis for Hot Rolling Simulation

6.1 Introduction.

In the present chapter the effect of hot rolling variables such as strain, strain rate, temperature and original grain size on kinetics of static recrystallization, recrystallized grain size and stress-strain behaviour are examined. The aim is to produce, as simply as possible, a set of parametric equations which can be used to describe:

a- static recrystallization

b- recrystallized grain size

c- stress-strain curves

in terms of the experimental variables studied. The set of equations established in this chapter can be used as a base for numerical hot rolling simulation of AISI316L.

6.2 Stress-strain Behaviour.

Samples were tested under plane strain compression at temperatures of 950, 1025, and 1100 C for two different grain sizes and a range of strain rates as shown in tables (6.1) and (6.2). The curves were then analysed with the view of obtaining the relationships necessary to describe the strength of the material as a function of the deformation variables and the original grain size. Some of the stress-strain curves obtained are shown in figure (6.1) for a range of the Zener-Hollomon parameter of four orders of magnitude. It can be seen that the strain to the peak is roughly constant and equal to 0.35 to 0.40 for the conditions tested.

The dependence on the strain rate, $\dot{\varepsilon}$, of the maximum value of the stress, σ_p, is shown in figure (6.2) from which the dependence of the strain rate on the inverse of the absolute testing temperature can be plotted as shown in figure (6.3). The slope of the lines in figure (6.3) is proportional to the activation energy for deformation whose value was found to be equal to 450kJ/mol. The dependence of the maximum stress on the original grain size is shown in figure (6.4) and detailed in table (6.2). It can be seen from figure (6.4) that σ_p increases slightly as the original grain size decreases for a constant value of the Zener-Hollomon parameter.

Figure (6.5) shows the dependence of the strain to the peak stress on the original structure of the material tested

under plane strain compression. It can be seen that the strain to the peak seems to be unaffected by the original grain size of the material for the conditions under which tests were carried out. The strain to the peak also seems to be insensitive to variations on Z, Zener-Hollomon parameter, as it can be seen from figure (6.6).

Figures (6.7), (6.8) and (6.9) show the dependence of the stress at a certain value of strain on the Zener-Hollomon parameter for a constant value of original grain size. Details are given in table (6.1). The stress at a particular strain level seems to be directly proportional to Z for the entire range of values studied of the Zener-Hollomon parameter. The dispersion obtained from experimental data seems to be largest in the case of the stresses at strain equal to 0.02. The data set available for each strain level was fitted by a straight line using the least square method. The equation set found is given by:

$$\sigma_{0.02} = - 565.3 + 36.0 \log_{10} Z \qquad (6.1)$$

$$\sigma_{0.1} = - 474.0 + 34.5 \log_{10} Z \qquad (6.2)$$

$$\sigma_p = - 506.0 + 37.0 \log_{10} Z \qquad (6.3)$$

$$\sigma_{ss} = - 410.2 + 30.3 \log_{10} Z \qquad (6.4)$$

Equations (6.1) to (6.4) were then used to predict the stress-strain curve for a certain value of the Zener-Hollomon parameter as it is shown in figure (6.10).

6.3 Kinetics of Static Recrystallization.

The static recrystallization kinetics were studied from samples tested under plane strain compression, axisymmetric compression or hot rolling. The results obtained were plotted as a function of the holding time before quenching according to the Avrami equation as depicted in figures (6.11) to (6.17) and detailed in tables (6.3) to (6.9). The experimental points were fitted by using the method of the least squares to give the best straight line. The slope of each line is k, the Avrami equation time exponent the values of which are indicated on each figure. The values of k vary from as low as ~0.2 to as high as ~1.5 for the data set presently available, although most of the values found were in the range of 0.5 to 1.0. It can also be noticed from figures (6.11) to (6.17) that k values tended to increase as either the initial strain applied to the sample increased or the original grain size decreased regardless of whether the sample was tested under plane strain compression or axisymmetric compression.

The time for 50% of material to recrystallize statically, t_{50}, is plotted against the equivalent strain applied to the sample tested under plane strain compression as shown in figure (6.18) and detailed in table (6.10). It can be observed that:

a- for each curve there is a critical value of strain, ε_c, above which t_{50} no longer is dependent on the strain applied to the sample.

b- the critical value of strain seems to be a function of the original grain size and the Zener-Hollomon parameter.

c- the critical strain is not less than 0.5 for the conditions shown in figure (6.18).

It can then be written that

$$t_{50} \; \alpha \; \varepsilon^{-3.6} \qquad \text{for} \qquad \varepsilon < \varepsilon_c \qquad\qquad (6.5)$$

The values of t_{50} are plotted against strain for samples tested under axisymmetric compression and hot rolled in figure (6.19). It is again clearly seen that there is a critical value of strain and its value is about 0.5. The t_{50} dependence on strain for axisymmetric samples at strains lower than the critical one appears to be similar to the one found for plane strain compression samples. There seems to be however, a slightly lower dependence of t_{50} on strain for the hot rolled samples.

The time for 50% of material to recrystallize statically is plotted against the inverse of the absolute testing temperature for two values of strain ($\varepsilon = 0.4$ and $\varepsilon = 2.0$) and two original grain sizes as shown in figure (6.20). The strain values were chosen such that the smaller was in the region of $\varepsilon < \varepsilon_c$ and the larger in the region of $\varepsilon > \varepsilon_c$. The slopes of the best straight lines fitted to the experimental points are a function of the apparent activation energy for recrystallization and can be used for its determination. The values of the apparent activation energy for static recrystallization found were:

$Q_{rex} = 475 \quad kJ/mol \qquad for \qquad \epsilon < \epsilon_c$

$Q_{rex} = 366 \quad kJ/mol \qquad for \qquad \epsilon > \epsilon_c$

The t_{50} dependence on the original grain size of the material deformed in the previous pass is shown in figure (6.21). The line shown in figure (6.21) represents a relationship of the form

$$t_{50} \; \alpha \; d_o^{1.33} \tag{6.6}$$

If it is assumed that the dependence of t_{50} on Z can be given by

$$t_{50} \; \alpha \; Z^{-0.38} \tag{6.7}$$

for any value of strain (Barraclough and Sellars,1979), then, it can be written that

$$t_{50} = A \; Z^{-0.38} \; d_o^{1.33} \; \epsilon^{-3.6} \; exp\left[\frac{Q_{rex}}{R \; T}\right] \quad ; \quad \epsilon < \epsilon_c \tag{6.8}$$

where

$Q_{rex} = 475 \pm 29 \; kJ/mol$

It can also be said about t_{50} that

$$t_{50} = B \; Z^{-0.38} \; exp\left[\frac{Q_{rex}}{R \; T}\right] \quad ; \quad \epsilon > \epsilon_c \tag{6.9}$$

where

$$Q_{rex} = 366 \pm 44 \text{ kJ/mol}$$

All available data from samples deformed at strains less than ε_c were plotted together as shown in figure (6.22). Points from axisymmetric compression as well as hot rolling samples were included in figure (6.22). Points from samples tested under plane strain compression were fitted by the best straight line. The constant of proportionality was found to be

$$A = 4.00 \; 10^{-15} \; s^{0.62} \; \mu^{-1.3} \tag{6.10}$$

Similarly, experimental points for strains higher than ε_c were plotted according to equation (6.9) as shown in figure (6.23). The constant of proportionality, B, was found to be

$$B = 3.00 \; 10^{-7} \; s^{0.62} \tag{6.11}$$

Equations (6.8) and (6.9) must produce the same value of t_{50} at strain $\varepsilon = \varepsilon_c$. Therefore,

$$\varepsilon_c = 6.64 \; 10^{-3} \; d_o^{0.36} \left[\exp\left[-\frac{Q''}{R\,T} \right] \right]^{-0.28} \tag{6.12}$$

where

$$Q'' = 109 \text{ kJ/mol}$$

6.4 Recrystallized Grain Size.

The static recrystallized grain size of samples tested under plane strain was plotted against the equivalent strain applied to the sample in the previous deformation as shown in figure (6.24) and detailed in table (6.11). There is a critical strain ε_* above which the recrystallized grain size becomes a function of the Zener-Hollomon parameter only. The critical strain is in the range of 0.6 to 1.0 for the conditions studied and it is apparently a function of Z as well as the original grain size of the material being tested. The recrystallized grain size is clearly dependent on strain for values of deformation lower than ε_* and this dependence does not seem to alter with the original grain size. It follows from figure (6.24) that

$$d_{rex} \alpha \; \varepsilon^{-1} \qquad \varepsilon < \varepsilon_* \qquad\qquad (6.13)$$

On the other hand, it can be observed from figure (6.24) that d_{rex} is independent of ε at strains higher than ε_*. Figure (6.25) shows the dependence of d_{rex} on Z for two values of pre-strain, nominally $\varepsilon = 2.0$ ($\varepsilon > \varepsilon_*$) and $\varepsilon = 0.5$ ($\varepsilon < \varepsilon_*$). The slope of lines in figure (6.25) gives the Zener-Hollomon parameter exponent. This appears to be independent of the strain applied to the sample. Therefore, it can be written that:

$$d_{rex} \alpha \; Z^{-0.12} \qquad\qquad (6.14)$$

Figure (6.26) illustrates the fact that d_{rex} is dependent on

the original grain size only if the pre-strain applied is less than the value of ε_*. In order to assess the dependence of d_{rex} on d_o, the recrystallized grain size must be plotted against the original grain size as shown in figure (6.27). The slopes of lines in figure (6.27) are 0.30, therefore,

$$d_{rex} \; \alpha \; d_o^{0.3} \qquad (6.15)$$

The experimental data available for samples with pre-strain less than ε_* were plotted according to

$$d_{rex} = A \; \varepsilon^{-1} \; Z^{-0.1} \; d_o^{0.30} \qquad \varepsilon < \varepsilon_* \qquad (6.16)$$

as it is shown in figure (6.28). Equation (6.16) seems to give good correlation between recrystallized grain size and variables studied. The constant A was found to be

$$A = 470.0 \; \mu^{0.7} \; s^{0.1} \qquad (6.17)$$

Limited experimental data is available for strains higher than ε_*. Nonetheless, plotting d_{rex} against $Z^{-0.1}$ according to

$$d_{rex} = B \; Z^{-0.1} \qquad ; \quad \varepsilon > \varepsilon_* \qquad (6.18)$$

produced

$$B = 2647.0 \; \mu \; s^{-0.1} \qquad (6.19)$$

Equations (6.16) and (6.19) must yield the same value of d_{rex} at strain equal to ε_*. Therefore,

$$\varepsilon_* = 0.18 \; d_o^{0.30} \qquad (6.20)$$

Figure (6.29) shows the dependence of the recrystallized grain size on the strain for samples hot rolled. The dotted line is the same dependence found for samples tested under plane strain compression.

6.5 Grain Growth.

Figure (6.30) shows the grain growth behaviour of AISI316L when deformed at 1100 C to a strain of 0.77 and at 1025 C to a strain of 0.39. Details are given in table (6.12). The full line in figure (6.30) was plotted from values calculated using equation (3.29).

6.6 Discussion.

6.6.1 Stress-strain Curves.

The general shape of a stress-strain curve obtained from an AISI316L stainless steel sample tested in plane strain compression can be sub-divided into three different sections:

a- a steep work hardening zone followed by

b- a shallow and wide maximum stress region and

c- a gentle transition into the steady-state part of the curve.

These characteristics are in overall agreement with most stress-strain curves reported for austenitic stainless steels (Barraclough,1974; Ryan et al.,1982; Al-Jouni,1983). Some differences arise when the values of strain to the peak stress, ε_p values, from different authors are compared. Usually, ε_p from torsion tests are higher than the ones reported from plane strain compression (Ryan et al.,1982). On the other hand, Towle and Gladman(1979) tested 316L steel at 1100 C and, $7.5s^{-1}$ to strains of ~0.6. Under these conditions, an ε_p value of ~0.3 to 0.4 is expected. However, the reported curve does not present any maximum stress.

The strain is not homogeneously distributed in a sample being deformed in plane strain compression (Sellars et al.,1976; Beynon,1979). If a nominal strain, ε_n equal to 0.33 is externally applied to a sample, one could expect internal local strain variations from ~0.1 (mid-width, ~1/5 thickness from tool/sample contact surface) to ~0.6 (at diagonal shear bands). If one assumes that:

    a- the critical strain for dynamic recrystallization to

     initiate, ε_c, is ~0.8 ε_p(Sellars,1980)

    b- the 'true' peak strain is the one reported for hot

     torsion samples (~0.5 at 1000 C)(Ryan et al.,1982)

then, the maximum nominal strain to be applied in a plane strain compression sample to start dynamic recrystallization is ~0.28. This is in agreement with double deformation tests carried out by Colàs(1983). Therefore, until ε_n~0.28 the sample should work harden only. Additionally, if the stress level at a given strain (ε<0.28) is determined by the highly deformed material, then the plane strain compression stress at a constant strain for 316 steel might be higher than the corresponding one for a 316 sample tested under, say, axisymmetric compression. This will be discussed later in chapter 7.

At strains between ~0.28 and ε_p, there will be some material dynamically recrystallized and some which will be dynamically recovered only. The dynamically recrystallizing material will pull the local stress value back to the expected one for an annealed material and an early peak stress appears as a result of this. The less deformed

material will only approach ϵ_c at strains as high as ~1.5 explaining the smooth transition into the steady-state stress regime.

6.6.1.1 Activation Energy for Hot Working.

The activation energy for hot working is in agreement with the average value found in the literature. The value of 450 kJ/mol determined is in accordance with the one reported by Hughes et al.(1974) of 460 kJ/mol. There are two other reported Q_{def} values for 316; the one by Ryan et al.(1982) (402 kJ/mol) and the one by Teodosiu et al.(1979) (499 kJ/mol). The value reported by Ryan et al.(1982) seems to be low for 316 steels since it has been often reported that Q_{def} for 304 steels is ~410 kJ/mol. As 316 steel is a more alloyed version of 304 steel, one would expect to have $Q_{def316} > Q_{def304}$. Moreover, the authors used an average stress, $\bar{\sigma}$, for assessing β and Q_{def} in the exponential law

$$Z = A'' \ \exp \ \beta\sigma_* \tag{6.21}$$

This should cause an over-estimation of Q_{def} rather than an apparently under-estimated value. This leads to another possibility, which is the presence of δ-ferrite in their experimental material. It has been recently pointed out that Q_{def} decreases considerably as the volume fraction of δ-ferrite in a dual phase austenitic stainless steel increases (Al-Jouni,1983). On the other hand, the higher value of Q_{def} found by Teodosiu et al.(1979) may be attributed to the

higher C content of the experimental steel used in their investigation ($C\sim0.55$ as opposed to the usual ~0.06 wt%). The value 460 kJ/mol was used to calculate Z values in the present research, regardless of the prior strain applied.

There has been some evidence that Q_{def} does not change significantly over a wide range of strains. Ahlblom(1977) verified that Q_{def} was approximately constant for 18-8 austenitic stainless steels at strains ~0.3 and at $\varepsilon \simeq \varepsilon_p$ (~0.5). There was some uncertainty however, as the strains were lower than ~0.15 since Q_{def} increased its value as ε decreased. Teodosiu et al. (1979) shows that Q_{def} varies only $\sim4\%$ as the prior strain varies from ~0.1 to ~2. However, they also reported an increase in Q_{def} as ε is lower than ~0.1. On the other hand, Foster(1981) testing Nb-steels has shown that Q_{def} can be considered constant since it varied from 396 kJ/mol to 406 kJ/mol as the strain increased from 0.2 to 0.4.

The constant activation energy for deformation, despite the change in the operative restoration process (from dynamic recovery only to simultaneous dynamic recovery and recrystallization with increasing strains) is useful since the stress can be calculated simply as a function of Z.

6.6.2 Stress-strain Curve Simulation.

The maximum stress in a sample being tested under plane strain compression is not significantly dependent on changes in the original grain size ($\sigma_{max} \propto d_0^{0.01}$). As previously mentioned in section(2.2), there is some uncertainty concerning σ_{max} behaviour when there are changes in the original grain size. One would expect that σ_{max} would increase as the original grain size decreases. The fact that this does not occur for 316L tested under plane strain compression suggests that some other variable (either a testing or material one) may be masking the strengthening action of refining the original microstructure.

As far as the strain to peak stress, ε_p, is concerned, one may generally write that

$$\varepsilon_p \propto d_0^n Z^m \qquad (6.22)$$

In the present work, n has been found to be ~0. This is in sharp disagreement with recent reports on austenitic stainless steels tested in torsion (Al-Jouni,1983). He found n~0.35 which coincides with the ones reviewed by Sellars(1980), n~0.5, for C-Mn steels. Most specimens in Sellars' review had been tested in torsion. Sellars and Al-Jouni's findings suggested that testing variables such as local strain, strain rate and temperature may have been producing a masking effect on the dependence of ε_p on d_0 when samples were tested under plane strain compression. However, Leduc(1980) found $\varepsilon_p \propto d_0^{0.3}$ for mild steel tested under plane strain compression.

On the other hand, the dependence of ε_p on Z is more clearly defined. The present work suggests that $\varepsilon_p \propto Z^{0.032}$. The Z exponent is fairly low when compared with values reported for C-Mn steels; 0.15 (Sellars,1980) and 0.17 (Leduc,1980). However, it is in agreement with the value recently reported for austenitic steels of m~0.08 (Al-Jouni,1983).

The expression finally adopted for stress-strain curve simulation was

$$\varepsilon_p = 0.097 \ Z^{0.032} \tag{6.23}$$

However, it is acknowledged that the relative roles that material as well as testing variables may have on the determination of ε_p. are not yet fully understood. Equations (2.5) to (2.11) together with equation (6.23) were used for simulation of stress-strain curves of 316 steel. A computer programme originally used by Leduc(1980) was slightly modified to account for the different material being used. A print-out of the programme is reproduced in appendix (6.1). The programme was run on a Hewlett-Packard desk-top computer model 9830A with plotter 9862A peripheral. This was very useful since interactive plotting could be performed.

The constants x, m, m' and k in equations (2.5) to (2.11) were essentially the same ones used in the original programme. An x value of 0.7 was used. This is lower than the average reported value (~0.79). However, changes in x from 0.7 to 0.8 does not significantly alter the stress-

strain predictions. The value used for m was 0.5, which is reasonable since the stress is expected to be proportional to the square of the dislocation density and this may vary linearly with the applied strain. There have been reports of experimental evidence supporting m' value of 1.4 for austenitic stainless steels (Roberts et al.,1979). The constant $k\sim0.5$ (0.49 in the original version) was kept the same. Finally, $\sigma_{ss(e)}$ equation was taken as σ_p. Previously, $\sigma_{ss(e)}$ equation was obtained by plotting hand drawn extrapolated stress-strain curves for mild steel as a function of Z (Leduc,1980). The fact of taking $\sigma_{ss(e)}$ as σ_p is a simplification for the problem. However, this does not seem to affect the model significantly and reasonable predictions of stress-strain curves were obtained for samples tested at temperatures in the range of 1100 to 950 C (figure (6.10)). As the temperature decreases, say to $\sim$800-850 C, the material will no longer recrystallize dynamically and the equations referred to above are no longer valid for simulation purposes. However, at 850 C, ε_p is $\sim$0.5 and ε_c $\sim$0.35. Therefore, the equations could still be used for hot rolling load calculations since it is unlikely that a slab will be given $\varepsilon = 0.35$ ($\sim$26% reduction) at such low temperatures.

6.6.3 Recrystallization Kinetics.

Studies of the kinetics of static recrystallization have been carried out at 3 different temperatures; 1100, 1025 and 950 C. Samples were tested at constant strain rate $\sim 5s^{-1}$, except for hot rolling specimens where $\dot{\varepsilon} = 3s^{-1}$. Three kinds of tests were used for kinetics studies: plane strain compression (most of the samples), axisymmetric compression and hot rolling. Plane strain compression was mainly carried out to produce data used in the derivation of equations (6.8) and (6.9). Hot rolling was used for comparison purposes whereas axisymmetric tests gave some information on the progress of static recrystallization in a sample more uniformly deformed.

6.6.3.1 Microstructural Behaviour.

Nucleation occurred mainly on grain boundaries although some intragranular nucleation was also present. This is in agreement with previous reports for austenitic stainless steels (Towle and Gladman,1979) and for α-iron (Glover and Sellars,1972). The freshly nucleated grain tended to grow along the existing grain boundaries and seldom on twin boundaries. Grains nucleated on grain boundaries grew faster than the ones nucleated on twin boundaries since the driving force for recrystallization and growth is higher at grain boundaries. Very often, a final recrystallized structure consisted of grains nucleated at grain boundaries which grew inwards to impinge on the

intragranularly nucleated ones, when these were present. It is worth mentioning that, although not frequent, intragranular nucleation happened in samples with original grain size 100μm. This contrasts with previous reports that this kind of nucleation only occurred in very large grains (~530μm) (Barraclough, 1974).

6.6.3.2 Recrystallization Rates.

The time exponent, k, in the Avrami equation (equation (3.5)) was found to be in the range of ~0.2 to ~1.5 for all the types of tests carried out. This agrees with reports on k values ranging from ~0.5 to ~1.5 for AISI316L (Towle and Gladman,1979). The lowest k value was obtained for plane strain compression test at 950 C, low strain range ($\epsilon < 0.4$) and samples with original grain size 260μm. The maximum k value obtained for the same test was ~1 at 1100 C, $\epsilon = 1.14$ and d_o 100 μm. Axisymmetric compression samples had k values always higher than 0.6 and the maximum k value was ~1.5. Hot rolling samples deformed in the range of $\epsilon = 0.2$ to $\epsilon = 0.5$ yielded k~0.8. Theoretically, if nucleation occurs mainly on grain boundaries, and the growth rate during recrystallization is kept constant, then k should be 1 (Cahn,1956). Reports of k lower than 1 for austenitic stainless steels have been frequently published. Towle and Gladman (1979) attribute the decrease in k to the fact that the growth rate during recrystallization is not constant and therefore Cahn's theory is not really

applicable. Other factors such as concurrent recovery (Capeletti et al.,1972) and influence of the prior dynamic softening mechanism on the static recrystallization (Glover and Sellars,1972) have also been reported to be important in causing a low value of k.

The k values ~0.4 to 0.2 obtained in the present work for plane strain compression samples may have resulted from the interaction of several different factors. At low strains ($\epsilon < 0.3$), the local strain in a plane strain compression sample is largely non-homogeneously distributed. When deformation is halted, there may be regions inside the sample which were deformed beyond an ϵ_c for dynamic recrystallization to initiate. Therefore, it is likely that dynamic recrystallization may have started locally. This region will recrystallize metadynamically once the deformation process has been completed. The material will recrystallize quickly, giving an initial high fraction recrystallized value in the early stages. Then, recrystallization will proceed into the less deformed regions of the sample. There static recrystallization may have already started or, accordingly to variables d_o, $\dot{\epsilon}$ and T, there may be a time lag in between the metadynamic and the static recrystallization process. In this process, the time for 50% of material to recrystallize may not be significantly affected. This will be discussed in more detail in chapter 7.

However, k may be considerably altered giving the low values measured in the present experiments. As the nominal strain

applied to a plane strain compression sample increases, the internal strain distribution becomes more homogeneous (Beynon, 1979). Therefore, the k value increases from ~0.5 to ~1 at 950 C for specimens 100μm original grain size. This increase in k with the nominal strain is consistently reproduced for all the other temperatures. However, the maximum k for plane strain compression samples is ~1, as opposed to ~1.5 for axisymmetric compression tests.

Axisymmetric compression samples have a more homogeneous strain distribution than plane strain ones, especially at low strains. The fact that k~0.6 for axisymmetric compression at $\varepsilon \approx 0.14$ and 1025 C supports the idea that the strain distribution is an important factor in determining k. It is interesting to see that k~0.7 has been recently reported for austenitic stainless tested under torsion (Al-Jouni,1983).

The grain size effect on k is clearly seen when curves from samples 260μm tested at 950 C and $\varepsilon \approx 0.3$ (k~0.2) are compared to specimens 100μm tested at 950 C and prior strain of $\varepsilon \approx 0.3$ (k~0.5). Clearly, this confirms reports that as d_o increases, k decreases (Barraclough,1974; Al-Jouni,1983).

In the present work, k was found to vary very little as the temperature changed (k~0.6; 0.58; 0.68 for samples deformed under plane strain compression to $\varepsilon \approx 0.4$ at 1100, 1025 and 950 C respectively). This confirms previous reports that k is insensitive to changes in test temperature (Wusatowski,1966; Glover and Sellars,1972; Barraclough,1974; Ahlblom,1977).

6.6.3.3 Effect of Strain.

The time for 50% of the material to recrystallize, t_{50}, is a strong function of the strain (t_{50} α $\epsilon^{-3.6}$) for strain less than the critical one, $\epsilon < \epsilon_*$. A strain exponent -4 has been frequently reported in the literature (Sellars,1980). The work hardening rate in a typical plane strain compression test starts to decrease at a critical strain for dynamic recrystallization, $\epsilon_c \approx 0.8\epsilon_p$, ~0.32. The stored energy for static recrystallization in the material being deformed decreases, since some regions of the specimen are recrystallizing dynamically. However, t_{50} decreases further with strain. At a strain well into the steady state, both structure and flow stress are independent of the strain applied (McQueen and Jonas,1975) and the stored energy for static recrystallization is constant. Therefore, t_{50} should only be dependent on the Zener-Hollomon parameter,Z. In fact, that has been verified by several workers (Glover and Sellars,1972; Djaic and Jonas,1972; Barraclough,1974; Towle and Gladman,1979). In the case of AISI316 tested in plane strain compression, ϵ_* is ~0.6 for samples of 100µm original grain size and ~0.8 for samples of 200µm grain size. The ϵ_* value is, for both samples, in between $\epsilon_p \approx 0.35$ and $\epsilon_{ss} \approx 0.7$ at 1100 C and ~1.0 to 1.5 at 1025 and 950 C.

If we assume that the strain distribution proposed by Beynon (1979) for plane strain compression samples deformed to ϵ_n ~0.33 is still applicable to $\epsilon_n \approx 0.6$, then it can be said that at least 50% of the material is dynamically recrystallizing at $\epsilon_n \approx 0.6$. This is in agreement with dynamic

recrystallization kinetics results followed by double deformation tests recently carried out in this department (Colás,1983). Therefore, it seems that a critical volume of material has to undergo dynamic recrystallization so that t_{50} becomes independent of strain. Curves at 1100; 100μm, 1025 and 950; 260μm suggests that there is an abrupt change of regime. This is in agreement with most results published on the low carbon steels (Morrison, 1972; Djaic and Jonas , 1972). However, reports on austenitic stainless steels suggest a smooth transition region from ε^{-4} dependence (Barraclough,1974; Barraclough and Sellars,1979; Towle and Gladman,1979).

It is important to note that ε_* remains fairly constant with temperature, but changes considerably with original grain size. This suggests that local strains at grain boundaries level may also have an important part to play in the determination of the critical volume of dynamically recrystallized material. Alternatively, local strains may be less homogeneously distributed on coarse grained samples leading to a bigger ε_*.

In contrast, $\varepsilon_* \approx \varepsilon_p$ for samples tested under axisymmetric compression. The strain in an axisymmetric sample is more evenly distributed than in plane strain compression. Therefore, when $\varepsilon \sim \varepsilon_p$ most of the sample may be undergoing nucleation of dynamic recrystallization. Ahlblom(1977) investigating dynamic recrystallization kinetics on AISI304 shows that at $\varepsilon \approx 0.6$ and $T \sim 1100$ C the fraction of material dynamically recrystallized is $\sim 25\%$ at least.

The t_{50} values measured from hot rolling samples are in agreement with the ones obtained from plane strain compression tests. There is a trend, however, for having shorter t_{50} for rolling samples at times longer than 1000s.

6.6.3.4 Effect of Grain Size.

The time for 50% of the material to recrystallize changes significantly with the original grain size ($t_{50} \propto d_o^{1.33}$). An exponent of 2 has been previously reported for the grain size dependence in 304 austenitic stainless steels (Barraclough, 1974). This was then used by Sellars(1980) to successfully correlate data on C-Mn steels from several other workers. However, Towle and Gladman (1979) studying 304 and 316 type steels, suggested that a d_o exponent lower than 2 may express a better dependence of t_{50}. The earlier review on the subject (section (3.6.3)) has indicated that change in the d_o exponent from 2 to 1 does not introduce a significantly better degree of correlation for the data examined. However, a d_o exponent of ~1.0 is consistent with the predominant grain boundary nucleation and growth observed in the present research. Since, the grain boundary surface per unit of volume, S_v, is inversely proportional to the grain size.

6.6.4 Recrystallized Grain Size.

The dependence of the recrystallized grain size on the equivalent strain previously applied to the sample follows a similar pattern to the one for recrystallization rate (figure(6.24)). The recrystallized grain size, d_{rex}, is strongly dependent on strain ($d_{rex} \propto \varepsilon^{-1}$) until a critical strain, ε_*, is achieved. This has a value ~0.6 for samples of original grain size 100 μm. The stored energy for recrystallization increases as the nominal strain on the sample increases. This, consequently, causes a decrease in the recrystallized grain size. For a constant stored energy, d_{rex} increases with test temperature since the nucleated new grains will grow faster.

A strain exponent -1 has been often reported in the literature (Sellars,1974; Barraclough,1974; Sellars,1980). However, a lower strain exponent (~0.5) has recently been reported to correlate d_{rex} experimental data on austenitic stainless types 304 and 316 (Towle and Gladman,1979). However, a review of several authors working with type 304 and 316 steels has indicated that a -1 exponent gives possibly a better correlation than -0.5 (see section (3.7)).

The dependence of d_{rex} on the Zener-Hollomon parameter, Z, (figure (6.25)) for strains lower than ε_* is in reasonable agreement with previously reported values for austenitic stainless steels (Barraclough,1974; Towle and Gladman,1979).

In a similar way to t_{50}, d_{rex} increases as the original

grain size increases (figure (6.26)). The d_O exponent of 0.3 (d_{rex} α $d_O^{0.3}$) is lower than the ones previously reported (0.5 to 1). Nonetheless, the lower d_O exponent gave a reasonable correlation for the experimental data available (Figure (6.28)).

The original grain size has a minor influence on the stress-strain curve level at strains lower than ε_* (figure (6.4)), although this might be caused by a non-homogeneously deformed sample effect. However, the same insensitivity of the stress-strain curve on d_O has been previously reported (Foster,1981; Al-Jouni,1983) for austenitic stainless steels. Therefore, it seems plausible to infer that d_O is most important in its role of providing available nucleation sites for recrystallization. If that is the case, an exponent ~1.0 would be the expected one for the range of d_O studied (~60 to 260μm). The fact that d_O exponent is lower than 1.0 suggests that not only the grain surface area per unit volume is the determining factor but also possibly intragranular nucleation and localized strains at grain and twin boundaries may have some effect. As the intragranular nucleation did not occur very often for the material conditions used, it may be that short range strain localization might have had a preponderant effect on d_{rex}. A lower exponent than 1.0 is consistent with observations by Barraclough (1974) for original grain sizes bigger than ~300μm for 304 austenitic stainless steels.

The critical strain, ε_*, has been reported to be $\approx 2\varepsilon_{ss}$ for ferritic stainless steels, where ε_{ss} is the strain

to the onset of the steady state (Glover and Sellars,1972).
In more recent review (Sellars,1980), it has been pointed
out that $\varepsilon_* \simeq \varepsilon_p$ for C-Mn and low alloy steels. In the present
investigation, however, it has been found that ε_* is in
between ε_p and ε_{ss}. This is consistent with t_{50} behaviour on
the equivalent strain and the same line of reasoning used
for that case may be applicable here.

Likewise, ε_* increases as the original grain size increases.
However, increasing the temperature has the opposite effect,
although changing the temperature from 1100 to 950 C does
not alter ε_* significantly (from ~0.6 to ~0.7) (figure
(6.26)).

As the sample is deformed further into the steady state
region, both stress and structure become independent of the
original grain size (Sah et al.,1974). The stored energy for
recrystallization remains constant and so does d_{rex}.

6.6.5 Grain Growth.

The grain growth after recrystallization is
negligible at 1025 C as seen in figure(6.30). At 1100 C, the
recrystallized grains grow faster. Nonetheless, for the
usual 15s inter-pass period of air cooling, a 45μm
recrystallized grain size would only grow to ~48μm. The
relatively slow grain growth rate of austenitic stainless
steels is consistently supported by previous workers
(Nörstrom,1974; Towle and Gladman,1979). Equation(3.29)
seems to describe suitably the experimental data available

(figure(6.30)). A grain size exponent lower than 10 has previously been pointed out as suitable for austenitic stainless steels (Towle and Gladman,1979).

An activation energy for isothermal grain growth is defined by equation(3.31) and its value is ~320kJ/mol. This is much lower than the activation energy for recrystallization (Q_{rex}~500kJ/mol) but very similar to the ones reported for creep in 304 (280 to 380kJ/mol) (Bhargava et al.,1976), in 316 (250±30kJ/mol) (Sritharan and Jones, 1981) and self-diffusion (260kJ/mol) (Assassa and Guiraldeng, 1978) for austenitic stainless steels.

The constant A" in equation(3.31) seems to yield a predicted value of grain size higher than the measured one when times are as long as 150s. Changing A" to ~$2x10^{-13}$ (~65% of its present value) gave a better correlation for the experimental data without altering significantly the predictions at short periods of time.

6.7 Conclusions.

1- The stress-strain curve of samples tested under plane strain compression has three distinct regions:

    a- steep work hardening part leading to a relatively early strain to peak stress, ε_p

    b- flat, wide maximum stress region

    c- gentle transition into the steady state.

2- The maximum stress is fairly insensitive to changes in

the original grain size from 60 to ⁻260µm.

3- The strain to peak stress, ε_p, is not a significant function of the original grain size.

4- ε_p varies only slightly with the Zener- Hollomon parameter ($\varepsilon_p \propto Z^{0.032}$).

5- ε_p can be adequately described by

$$\varepsilon_p = 0.097 \ Z^{0.032}$$

for the range of conditions studied.

6- The distribution of strain in a sample being deformed under plane strain compression is an important variable in the determination of the overall shape of the stress-strain curve for type 316 stainless steel.

7- An exponential law ($Z = A'' \exp \beta\sigma$) is applicable for σ_0, $\sigma_{0.1}$, σ_p, $\sigma_{ss} > 100 MPa$.

8- Static recrystallization occurs predominantly by grain boundary nucleation. Large original grain sizes show isolated intragranular nucleation.

9- Recrystallization follows an Avrami type behaviour with the time exponent k within 0.2 and 1.5.

10- The strain distribution inside a sample being tested under plane strain compression is an important variable in the determination of k.

11- The value of k increases when the applied strain increases. However, k decreases as the original grain size increases and it does not significantly change when the temperature is altered from 1100 to 950 C.

12- The time for 50% of material to recrystallize is

shortened by increasing the applied strain. A strain exponent -3.6 is applicable at low strains, however at the steady state this becomes 0.

13- The two different dependences of t_{50} on the strain shows an abrupt break between them at a critical strain ε_*.

14- ε_* increases significantly with the original grain size. On the other hand, it changes little as the test temperature changes from 1100 to 950 C.

15- ε_* has a value in between ε_p and ε_{ss}, the strain to the onset of the steady state, for samples tested under plane strain compression. However, $\varepsilon_* = \varepsilon_p$ for samples tested under axisymmetric compression. An attempt to explain this behaviour has been made in terms of differences in strain distribution within samples tested under different modes of deformation.

16- An equation of the type

$$t_{50} \alpha Z^{-0.38} \varepsilon^{-3.6} d_o^{1.33} exp\left[\frac{Q_{rex}}{RT}\right]$$

gave reasonable correlation for t_{50} obtained from plane strain compression and most experimental data obtained from other tests when $\varepsilon < \varepsilon_p$.

17- The recrystallized grain size dependence on the equivalent strain follows a pattern similar to the t_{50} dependence:

a- $d_{rex} \alpha d_o^{0.3} \varepsilon^{-1} Z^{-0.1}$ when $\varepsilon < \varepsilon_*$

b- d_{rex} is not a function of strain for $\varepsilon > \varepsilon_*$

c- ε_* is in between ε_p and ε_{ss}

18- The isothermal grain growth of AISI316L can be described by

$$d^2 = d^2_{rex} + kt$$

19- The grain growth is not significant at short times (<15s) even at high temperatures (1100 C).

CHAPTER 7

Strain Distribution Effect on Static Recrystallization

7.1 Introduction.

Figure (7.1) shows the strain distribution on the longitudinal cross section of a specimen tested to a strain of 0.345 under plane strain compression (Beynon,1979). There are regions of high values of deformation (ε = 0.6 to 0.8) in which the local strains are roughly twice the value of the applied external strain. However, there are regions in which the strains applied locally are half or one third of the external strain to which the sample is being subjected. The main purpose of this chapter is to examine the effects that the inhomogeneous strain distribution may have on the kinetics of static recrystallization for samples tested under plane strain compression.

7.2 Kinetics of Static Recrystallization.

Samples tested under plane strain compression were sectioned at line A'A' as shown in figure (7.1) and detailed in table (7.1). The fraction recrystallized was then measured at lines perpendicular to the plane of figure (7.1). The trace of these lines can be seen from figure (7.1) represented by points A, B and C. It was expected that at point A the local strain would be greater than 0.4, at B between 0.3 and 0.4 and at point C less than 0.3. Figure (7.2) shows the kinetics of static recrystallization at lines whose traces are A, B, and C on samples tested as shown in table (7.1). The dotted lines are the ones which are obtained from a constant strain value according to figure (6.18) using an Avrami equation time exponent of 1. It can be concluded from figure (7.1) that:

a- The kinetics behaviour at lines A, B and C is in agreement with what is expected as shown by the dotted lines derived from figure (6.18).

b- The values of k found for lines A, B and C are equal to 1

The total fraction of material statically recrystalllized can be obtained if the parts correspondent to each value of strain are added together at a given annealling time as detailed in appendix (7.1). In this way a curve can be built as shown in figure (7.3) (full line) and compared with the one obtained experimentally (dotted line). It can be seen from figure (7.3) that the value of k, the Avrami time

exponent, changes slightly from 0.57, dotted curve, to 0.69, full line. However, the time for 50% material to recrystallize, t_{50}, does not change significantly.

Figure (7.4) shows the fraction recrystallized versus time dependence for curve B of figure (7.2). The values of dX/dt, the recrystallization rate, can be then read out for a constant value of fraction recrystallized and used in

$$G = \frac{dX/dt}{A} \qquad (7.1)$$

where A is the migrating grain boundary area and G is the recrystallizing grain growth rate. In this manner, the growth rate of the recrystallizing grains is calculated. Curve B was chosen for the measurements because it has the best spectrum of fraction recrystallized values from the data set analysed. The migrating grain boundary area measured at lines of constant strain for the samples from curve B in figure (7.2) are shown as a function of the fraction recrystallized in figure (7.5).

Figure (7.6) shows the recrystallizing grain growth rate dependence on the annealing time for regions of constant strain of samples used to built curve B in figure (7.2).

7.3 Restoration Kinetics.

Samples were tested under plane strain compression at constant temperature and strain rate. Two deformations were given with a resting time interval in between them. The

samples were kept inside the furnace during the whole of the test. The results from double deformation tests were then used to assess the restoration kinetics of the material under hot working conditions. A restoration index as defined by equation (5.24) and schematically depicted in figure (7.7) was calculated after each test. Figures (7.8), (7.9) and (7.10) show the stress-strain curves for samples with original grain size 100μm double deformed at temperatures of 960 and 1033 C to strains of 0.33, 0.90 and 0.29. Figure (7.11) shows the fraction restored dependence on the inter-pass annealing time. Details of test conditions are given in table (7.2).

7.4 Static Recrystallization Model.

7.4.1 Introduction.

The fraction recrystallized of a sample deformed under plane strain compression was measured on the deformation zone. The term deformation zone refers to the material which was in between the pair of tools when the test was completed. It is known from figure (7.1) that the strain distribution in this region is highly non-uniform. Therefore, the kinetics of recrystallization which takes place during resting periods may be affected by this non-uniformity of strain. It is consequently of interest to know

how the recrystallized fraction measured on the entire gauge longitudinal cross section of the sample can be compared with the fraction recrystallized measured from a sample uniformly deformed. Both samples, the one deformed uniformly and the one deformed non-uniformly must have been subjected to an identical external strain. Particular interest is drawn to the value of t_{50} and the effects that strain distribution may have on its uniform value.

7.4.2 The Model.

The longitudinal cross section of the samples was subdivided into small rectangular elements as shown by the gridded quadrant in figure (7.1). The actual size of each element is 0.5x1mm for a 7x7 element grid superimposed to a sample deformed to $\varepsilon = 0.345$. A strain value was assigned to each element in the following manner:

a- first the strains at each element corner were added and averaged in order to generate a strain value ascribed to a certain element in a given quadrant.

b- then each element strain value was added to its symmetrical counterpart from other quadrants. An average value was then obtained.

A strain distribution matrix for the sample was obtained in this manner.

At any time time t, the fraction recrystallized in an element A_{ij} could be calculated using expressions (6.8) and (6.9). The assumption was made that each element would

recrystallize statically following an Avrami type equation with a time exponent equal to 1. Therefore,

$$x_{ij} = 1 - \exp\left[-3.0 \; \frac{t_{ij}}{t_{95}}\right] \qquad (7.2)$$

where x_{ij} is the fraction recrystallized of the element ij at a time t. Accordingly,

$$t_{95ij} = 4.32 \; t_{50ij} \qquad (7.3)$$

where t_{50ij} is the time for 50% of element ij to recrystallize and t_{95ij} is the time for 95%.

The total volume of material to recrystallize, v_r, can be written as

$$v_r = X \; V \qquad (7.4)$$

where X is the fraction of material recrystallized measured on the longitudinal cross section of the sample whereas V is the total volume of material undergoing static recrystallization.

Another form for expressing v_r is

$$v_r = \sum_1^n \sum_1^m v_{rij} \qquad (7.5)$$

where v_{rij} is the volume of recrystallized material in an element ij. Hence,

$$v_r = \sum_1^n \sum_1^m x_{ij} \; v_{ij} \qquad (7.6)$$

If all the elements have the same volume, then

$$v_{ij} = \frac{V}{m\ n}$$

and it follows that

$$X = \frac{1}{m\ n}\sum_{1}^{n}\sum_{1}^{m} x_{ij} \qquad (7.7)$$

Equation (7.7) implicitly assumes that there is no interrelation between elements as far as the kinetics of static recrystallization is concerned. Therefore, each element may be considered as an independent cell unit. It is also assumed that the amount of softening occurring as a result of static recrystallization inside a given element is additive. The total sum gives the value of fraction recrystallized for the whole sample.

The maximum migrating grain boundary area dependence on the strain was taken from figure (7.12)(Towle and Gladman,1979) from which it can be written that:

$$Smax_{ij} = 13.78 + 7.19\ \varepsilon_{ij} \qquad (7.8)$$

and ε_{ij} stands for the equivalent strain at an element ij. The dependence of the migrating grain boundary area on the fraction recrystallized can be derived from figure (7.5) as being

$$S_{ij} = 4\ Smax_{ij}\ x_{ij}\ (\ 1 - x_{ij}\ ) \qquad (7.9)$$

Therefore, an average migrating grain boundary area can be calculated from the elemental ones using

$$S = \frac{1}{m\,n}\sum_{1}^{n}\sum_{1}^{m}S_{ij} \qquad\qquad (7.10)$$

The average grain growth rate over the longitudinal cross section of the sample is then calculated from

$$G = \frac{dX/dt}{S} \qquad\qquad (7.11)$$

The set of equations above described was used in a computer programme which is described in detail in appendix (7.2).

7.4.3 Results.

The fraction recrystallized dependence on annealing time for a material deformed at 1033 C to a strain of 0.345 is shown in figure (7.13). It can be observed from figure (7.13) that the elements on the diagonal recrystallize faster than the expected for the sample as a whole. It can also be seen that the homogeneously deformed sample takes less time to recrystallize 100% than the sample inhomogeneously deformed although the latter seems to have started to recrystallize first. There is no significant change in the value of t_{50} due to strain distribution. An inhomogeneous strain distribution apparently affects the k value of the Avrami exponent as shown in figure (7.14). In this figure, the fraction recrystallized is plotted against time according to an Avrami equation. It can be noticed from figure (7.14) that k approaches the assumed value of 1 for

the diagonal elements. The inhomogeneously deformed sample kinetics of static recrystallization does not seem to obey an Avrami type of equation. Any attempt to fit a straight line to the data for the inhomogeneously deformed sample would lead to a value of k lower than 1.

The effect of an inhomogeneous distribution of strain on the average migrating grain boundary area behaviour is shown in figure (7.15). There appears to be a shift of the maximum of the curve for the average migrating grain boundary towards a lower than 50% value of fraction recrystallized. Figure (7.16) shows the grain growth rates for an average inhomogeneously deformed sample and for its diagonal elements. The grain growth rate of a sample uniformly deformed is also shown in figure (7.16) for comparison purposes.

Figures (7.17),(7.18) and (7.19) show a comparison between restoration function calculated from the model and the experimental values of restoration index obtained from double deformation tests. There is reasonable agreement between experimental and predicted values for cases in which samples were deformed to strains of 0.345. The agreement is less successful for samples deformed to strains of 0.9 as it can be seen from figure (7.19). However, in computing theoretical values plotted in figure (7.19), the assumption was made that the strain distribution was similar to the one found for $\epsilon = 0.345$. The strain in each element was varied then proportionally to the external value of strain applied to the sample.

7.5 Discussion.

7.5.1 Kinetics of Static Recrystallization.

The nucleation process at the start of static recrystallization in 316L austenitic stainless steels happens mostly on grain boundaries. Intragranular nucleation also occurs though only very rarely. Cahn(1956) has pointed out that in such situations the Avrami equation time exponent, k, should be equal to 1. Measurements of fraction recrystallized at lines of constant strain (figure (7.2)) indicates that k ~1. This suggests that k values less than 1 may only have resulted from an effect of local strains on the kinetics of static recrystallization.

If a strain distribution like the one shown in figure (7.1) is assumed to be the true one for the sample being deformed, then a simple calculation based on the percent of material deformed to a given extent may produce figure(7.3). It is clear from this sample that the assumed ε distribution describes suitably the local deformation of the sample. Additionally, figure(7.3) clearly shows that an apparent k value less than 1 may result from local inhomogeneous strain distribution. In fact, the most deformed regions will recrystallize faster and before any other inside the sample. The less deformed regions of the sample will follow the recrystallization process. These waves of local static recrystallization may overlap each other as the annealing

process takes place. Alternatively, the waves may be separated from each other by a given period of time. This, clearly, will be dependent on the magnitude of <u>local</u> hot working variables. Whatever the particular kinetics, the net macroscopic effect will be a decrease in k value. Therefore, at short periods of time, the measured fraction recrystallized will be higher than the expected one for a homogeneously deformed sample. On the other hand, at long periods of time the fraction recrystallized will be lower than the expected. However, it remains to be clarified what is the effect of local strains on the time for 50% recrystallization.

The grain growth rate during static recrystallization decreases with time (figure(7.6)). The experimental points can be fitted by an expression of the type

$$G \propto t^{-0.8}$$

The exponent 0.8 is slightly higher than the previously reported values of 0.7 for α-iron (Glover and Sellars,1972) and for austenitic stainless steels (Towle and Gladman,1979). There have been several reports in the literature supporting a decreasing growth rate with annealing time. Wusatowski (1966) explains this effect in terms of a decrease in driving force for recrystallization due to static recovery. English and Backofen (1964) suggested that a segregation phenomenon ahead of the grain boundary may have caused the decrease in growth rate with

time. Glover and Sellars (1972) showed that the previous dynamic softening mechanism may cause the growth rate to be constant or a function of time. Concurrent recovery has also been used to explain the growth rate behaviour observed in austenitic stainless steels type 316 and 304 (Towle and Gladman,1979). These authors have also pointed out that a solute drag effect is unlikely to happen since the time exponent in the Avrami equation increased as the temperature decreased. In the present work, the growth rate has been measured at a single temperature. However, the k value does not seem to have altered with test temperature. This may indicate that a solute drag explanation for the decreasing growth rate with time can be ruled out.

The nominal strain applied to the sample was ~0.30. The local strain at line B was constant and ~0.30 which is of the order of the critical strain for dynamic recrystallization. Therefore, few, if any, nuclei should be expected to be present before static recrystallization has started. For this reason, it is thought that the concurrent effect of recovery on the static recrystallization may have caused the observed dependence of G on the annealing time.

Static recrystallization kinetics was followed using double deformation tests (figure (7.10)). The main characteristic of this kind of test is that the time for 50% of material to soften appears to be shorter than the ones expected from direct metallographic measurements. At 950C and ε~0.3, the t_{50} from double deformation tests is ~50s, whereas the t_{50} from metallographic observations is ~400s

(see figure (6.18)). From this figure, it can be seen that a t_{s_0}~50s corresponds to a sample pre-deformed to $\dot{\varepsilon}$~0.6. However, 0.6 is the approximate value of strain expected to occur in the diagonal of the plane strain compression sample. An explanation for this behaviour could be the high deformation zone existent at the sample diagonal. As soon as the deformation process halts, the diagonal region would recrystallize whereas other parts of the sample would be recrystallizing or still in an incubation period. When a second deformation is given, it is likely that the diagonal region will be the softer part of the material, accommodating most external strain applied to the sample. Therefore, using the double deformation technique for measuring kinetics of static recrystallization could lead to spurious values of fraction recrystallized and, consequently, false values of t_{s_0}. In fact, what could be measured would be the static softening kinetics in the regions of the sample where the strains are highly localized.

A sample tested at ~950C and ε~0.3 yielded t_{s_0}~10s. However, figure(6.18) shows that a t_{s_0}~10s is the expected one for a sample ε~0.6. This behaviour is quite similar to the one found for 1025C.

Another variable that may be considered is the local increase in temperature due to adiabatic heating. All samples in figure (7.10) were tested at ~5s^{-1}. Therefore, a local strain of ~0.6 will be achieved in ~0.1s which is a very short period of time compared with the time needed for

heat conduction to take place to the sample surroundings.
However, a maximum increase in temperature of the order ~20C
is expected in the zones of more deformation when the sample
is externally deformed to ~0.3. The adiabatic heating factor
becomes more important when the strain is increased. The
sample deformed to a nominal strain of 0.9 presents a t_{s_0}
value of ~1.5s. The t_{s_0} value calculated from figure(6.18)
suggests a t_{s_0} of ~7s. Any increase in the strain in the
100μm/1025C curve shown in figure (6.18) will not decrease
t_{s_0} any further. Therefore, adiabatic heating may be held
responsible for this extra decrease in t_{s_0} value. In fact,
an increase in local temperatures of ~75 C (1100C curve)
could account for a decrease in t_{s_0} from 7s to ~2s.

The strain at the sample diagonals will be ~1.8. This will
be applied in a period of time of ~0.6s. Figure(6.1) shows
that the stress level at 1025C is ~200MPa . Therefore, the
energy per unit volume imparted to the sample is
~360x10^6 N/mm^2. The density and specific heat at 1025C are
7688 kg/m^3 and 658 J/kg C respectively. Therefore, the
expected increase in the sample temperature at the diagonal
shear band region is of ~71 C. It is, therefore, likely that
the most deformed regions of the sample may achieve an
increase in temperature of ~70 C. Recent measurements of
diagonal temperature in plane strain compression tests
carried out in 316L stainless steel at 1000 C indicated an
increase in ΔT of ~70 C (Colás,1983), confirming previous
calculations.

In view of the above discussion, it can be concluded that

local strains and adiabatic heating may produce apparent softening kinetics from double deformation tests which do not reflect the results obtained from metallographic measurements.

7.5.2 Static Recrystallization Model.

The modelling for static recrystallization kinetics in a sample deformed non-homogeneously in plane strain compression was developed as an attempt to answer important questions raised in the early stages of the present investigation. The main points of attention were:

a- The effect of the strain distribution on the time for 50% recrystallization

b- A comparison of the static recrystallization kinetics of the elements in the diagonal with the one measured from double deformation tests.

In figure (7.13), the fraction recrystallized versus annealing time is shown for a sample uniformly deformed to $\varepsilon \sim 0.3$, for a sample non-uniformly deformed and for its diagonal elements only. The Avrami equation time exponent for the homogeneous deformed sample was assumed to be 1. Clearly, the non-homogeneous sample has a higher value of recrystallized fraction during the early stages of recrystallization. On the other hand, the sample homogeneously deformed finishes its recrystallization process well ahead of the non-homogeneous one. However, the time for 50% of material to recrystallize for both samples

is nearly the same. This may indicate that t_{50} is not significantly affected, although k values are. The implication of this result is that kinetics of static recrystallization in plane strain compression tests measured on an overall sample gauge area can be used in the assessment of t_{50} values. However, k values obtained from these curves may be severely affected by the localization of deformation.

The transformation curve for diagonal elements is nearly parallel to the one for homogeneous samples. This is an indication that the strain distribution in the diagonal elements is more uniform than for the rest of the sample.

The curves from figure(7.13) were re-plotted as shown in figure(7.14). In this way, k values for samples non-homogeneously deformed and its diagonal elements could be assessed. The curve for the whole sample cannot be fitted by a straight line. Therefore, an Avrami equation is not applicable to describe the process. This is supported by similar findings reported in the literature (Campbell et al.,1974, Towle and Gladman,1979). In measuring fractions of material recrystallized, it is likely that X will be no less than ~10 to 15% and no more than ~80 to 90%. For this particular band of X values, the k value expected would be ~0.6. However, given the nature of the fraction recrystallization measurement methods, one would not detect that the curvature existed. Alternatively, the deviation from a straight line could be seen by the observer as a result of experimental scatter.

At the present stage, one could comment on the limitations inherent to the model used in the simulations. Perhaps the most important of all is that the model does not take into account an increase in local temperature due to plastic deformation. The heat accumulated during deformation would dissipate during the annealing period. Consequently, its effects are likely to be more pronounced in the early stages of the recrystallization. An increase in local temperature accelerates the rate of transformation re-enforcing the strain distribution effect. The net result would be an even lower value of k than the ones predicted in figure (7.14).

Another limitation in the present model is that it assumes a homogeneous strain distribution in a grid element. This is $\sim 0.5 \times 1.0$ mm^2 for a 7x7 element grid used in the simulations carried out in this chapter. It is unlikely that this is a realistic assumption. This becomes more evident when simulation for two samples deformed under the same conditions but with different original grain sizes is performed. Clearly, a change in d_o implies a corresponding change in t_{so}. However, it has been experimentally observed that k decreases as the original grain size increases. Nonetheless, the model does not predict any change in k value with the original grain size. This suggests that the strain distribution may change with the original grain size, becoming more non-homogeneous as d_o increases.

Nevertheless, the model predicted kinetics of recrystallization is in reasonable agreement with the one

measured from samples double deformed (figures(7.17) and (7.18)). In these cases the strain externally applied to the sample is ~0.3 and an increase in local temperature, though 20C, is not significant. Moreover, the interpass times are from ~10 to 100s causing the adiabatic heat effect to be minimized.

This, however, is not the case for the sample deformed to ε~0.9 and as a result the predicted kinetics is in disagreement with the measured ones. In this instance, adiabatic heating can explain the observed discrepancies between predicted and observed kinetics of static recrystallization.

7.6 Conclusions.

1- The values of k are strongly dependent on the distribution of local strain in samples tested under plane strain compression. On the other hand, k values measured from recrystallization curves obtained from data measured on iso-strain lines produced k~1.

2- The grain growth rate during recrystallization on iso-strain lines decreased with the annealing time. This is possibly due to the decrease in driving force for recrystallization as a result of concurrent recovery.

3- Kinetics of static recrystallization measured from double deformation tests is in disagreement with the one measured from direct metallographic observation. The cause of this

disagreement is thought to be the additive effects of local strain distribution at the most deformed zones of the sample and adiabatic heating.

4- The time for 50% recrystallization does not alter significantly as a result of the non-homogeneous strain distribution.

5- The kinetics of static recrystallization of a sample deformed non-homogeneously under plane strain compression cannot be fitted by an Avrami equation.

CHAPTER 8

Hot Rolling Strength and Simulation of AISI316L

8.1 Introduction.

AISI316L, the material used for the present hot rolling simulation studies, was prepared as small slabs 100x50x25mm original nominal geometry. Several hot rolling schedules were carried out as described in chapter 5.

The three main dependent variables analysed in the present chapter are :

    a - average mean plane strain strength

    b - temperature evolution during air cooling

    c - structural behaviour of the material.

The last item can be split into two sub-sections; one concerning the dependency of the fraction recrystallized on the mechanical processing variables and the other involving the study of the average recrystallized grain size. Theoretical predictions about microstructure and strength are made using a modified version of Leduc's computer programme (Leduc,1980). In addition, material and information on hot rolling schedules are generated for further simulation work using plane strain compression tests

as reported in chapter 9.

8.2 Original Structure.

In order to know the original structure to be hot rolled, some samples were re-heated at temperatures ranging from 900 to 1250 C for constant periods of time (1/2 hour). The samples were quenched and their grain size measured. Figure (8.1) shows how the original grain size changes as a function of the re-heating temperature. The original grain size remains reasonably constant for temperatures ranging from 900 to 1150 C. From 1150 C onwards, there is an increase in the original grain size within a narrow band of re-heating conditions. The original grain size can be as large as 300 μm after re-heating for 1/2 hour at ~1270 C.

8.3 Temperature Evolution During Hot Rolling.

It was initially thought that the AISI316L type steel would have the same cooling characteristics as for AISI304. Figure (8.2) shows that this is the case for air cooling periods shorter than 20s. Air cooling behaviour for both steels differs markedly for air cooling periods longer than 20s. AISI316L steel cools faster than AISI304 of same geometry. The difference between cooling curves can be as much as 30 C after 100s air cooling.

Figure (8.3) shows AISI316L air cooling behaviour when kept in air held by tongs or resting on the points of three nails. There are no significant differences between both cooling curves for periods of time as long as 180s. Either temperature versus time profile can be used for the calculation of the heat transfer coefficients during air cooling.

The heat transfer coefficient dependence on the surface temperature of a slab being held by tongs is shown in figure(8.4). Details of the procedure used for the heat transfer coefficient calculations as well as detailed discussion of the computer programme used is presented in appendix(8.1).

The curve which best fitted the experimental data available is given by

$$H_{theor} = -0.0368(\ T_s - RT\ ) - 0.4579 \times 10^{-10}(\ T_s^4 - RT^4\ ) \quad (8.1)$$

where the symbols mean

H_{theor} = theoretical surface heat transfer coefficient, represented by a full line curve in figure(8.4)

T_s = slab surface temperature, in degrees K

RT = room temperature, taken as 298 K

The emissivity was found to be equal to 0.81. Figure(8.5) shows a comparison between an experimental air cooling curve and a theoretical one calculated using equation(8.1). It can be observed from figure(8.5) that experimental air cooling behaviour can be reasonably simulated when equation(8.1) is used.

8.4 Spread During Hot Rolling.

A specimen being hot rolled is mainly deformed in the longitudinal and vertical directions. There is however, a finite lateral deformation which occurred during hot rolling of all experimental slabs processed. The samples initially had widths of either ~51mm or ~25mm. A slab 51mm initial width can be as wide as 54.5mm after undergoing a typical schedule of three passes 25% reduction in thickness each. A typical final width value for samples 25mm original width after processing under the same schedule as for 51mm samples was ~28mm. Therefore, a wide sample can spread as much as 7% whereas a narrow one as much as 12% when deformed from an original thickness of ~25mm to 10.8mm final thickness. The intermediate widths during a multi-pass schedule are usually unknown. Hence, it is necessary to calculate the intermediate pass width values since they are used in the computation of the mean plane strain stress for each rolling pass. .

Beese's equation (Beese,1972) for predicting spread during hot rolling is

$$S = A \left[\frac{h_i}{w_i} \right]^B \exp - C \frac{h_i}{\sqrt{R \, \Delta h}} \qquad (8.2)$$

where

$$S = \frac{\ln \left(w_{i+1}^{max} / w_i \right)}{\ln \left(h_i / h_{i+1} \right)}$$

and w^{max} is the maximum width. The other symbols have their usual meaning.

A, B and C are constants with values 0.61, 1.3 and 0.32 respectively. If the original sample geometry is kept constant, equation(8.2) may be re-written as

$$S = D \, exp\text{-}C \, \frac{h_i}{\sqrt{R \, \Delta h}} \qquad (8.3)$$

The mean final width as defined by equation(5. 6) was used in the calculations of the experimental spread values. The available experimental points were plotted as shown in figure(8.6) for 51mm width samples and figure(8.7) for samples 25mm original width. A best straight line was fitted to the data set, and the resulting equations are

$$S = 0.02479 \, exp \, 1.265 \, \frac{h_i}{\sqrt{R \, \Delta h}} \qquad ; w_i \approx 51mm \qquad (8.4)$$

and

$$S = 0.054 \, exp \, 1.336 \, \frac{h_i}{\sqrt{R \, \Delta h}} \qquad ; w_i \approx 25mm \qquad (8.5)$$

The spread can be expressed as

$$S = \frac{\ln (\bar{w}_{i+1} / w_i)}{\varepsilon'} \qquad (8.6)$$

where ε' is the total plane strain undergone by a sample being hot rolled. Therefore,

$$\bar{w}_{i+1} = w_i \, exp (\varepsilon' \, S) \qquad (8.7)$$

may be used for predicting intermediate widths during a multiple pass hot rolling schedule.

Figure(8.8) shows a comparison between experimental mean width values and theoretical ones obtained using equation (8.6) for samples 51mm original width. Figure(8.9) shows the same correlation for samples 25mm original width.

The relative error computed as

$$RE\% = 100 \cdot abs \left[\frac{\bar{W}_{exp} - \bar{W}_{theor}}{\bar{W}_{exp}} \right] \qquad (8.8)$$

appeared to be no more than 3.5% for both wide and narrow samples. Most of the predicted values had a relative error less than 2%.

8.5 Strength During Hot Rolling.

8.5.1 Effect of Original Microstructure and Rolling Reduction.

Figure(8.10) shows the dependence of the mean plane strain strength on the original sample microstructure during hot rolling. All samples had the same original thickness (10.8mm). In order to obtain samples with 30 μm original grain size, specimens were re-heated before test at

960 C for 15 minutes.

The value of the mean plane strain strength remains approximately constant and equal to 275 MPa for samples with an original grain size coarser than 60 µm. However, for samples with finer grain sizes, there is a significant dependence of the mean plane strain strength on the original structure. The onset of this dependency seems to be within the range of 40-60 µm original grain size for the conditions of hot rolling used in the present experiments. The load dependence on the reduction given in a rolling pass is shown in figure(8.11). The dependence of the mean equivalent stress on the rolling reduction is shown in figure(8.12). The rolling load rises steadily as a function of the reduction given to the material. However the mean flow stress rises continually until a certain value of strain applied to the sample is achieved. From there onwards, the mean flow stress keeps apparently constant as the rolling reduction increases.

8.5.2 Effect of Strain Rate and Rolling Temperature.

The dependence of the mean plane strain strength on the Zener-Hollomon parameter is shown in figure(8.13). Each sample was given 25% reduction in a single pass rolling schedule. All samples had an original thickness of 25mm and finished with a 18.8mm final thickness. Four samples were taken out of the 18.8mm thickness batch and heat treated to give an original grain size equal to 100 µm. The samples

were then rolled a second time with a single pass of 25%
reduction and quenched immediately. The samples' final
thicknesses were 14.1mm. The same route was repeated so that
a 14.1mm initial thickness sample was re-heated and rolled
for a third time in a single pass rolling schedule. In this
manner, it was possible to study the effect of the original
geometry on the rolling mean plane strain strength. As it
can be observed from figure(8.13), the initial geometry does
not appear to affect the hot rolling strength significantly
for the conditions presently studied.

The experimental data set available can be correlated by an
exponential type of equation for all Z values. The equation
which represents the best straight line fitted to the
experimental points is given by

$$\bar{\sigma}' = -322 + 28.3 \log_{10} Z \qquad\qquad (8.9)$$

where $\bar{\sigma}'$ is the mean plane strain strength.

Figure(8.14) shows the dependence of the mean
plane strain strength on Z for second pass samples deformed
in 3x25% rolling schedule. Some of the samples had a
partially recrystallized structure when entering the pass.
Others had a fully unrecrystallized structure. The dotted
line in figure(8.14) represents equation(8.9). Samples with
partially recrystallized microstructure have lower strength
than the ones completely unrecrystallized prior to the
second pass being given. The differences between
unrecrystallized sample strength and first pass sample
strength is ~40 MPa for the range of Z studied. This is the

same as to say that

$$\bar{\sigma}'_{2unrex} \simeq 1.15 \; \bar{\sigma}'_1 \tag{8.10}$$

where the subscripts stand for

    2unrex = second pass, unrecrystallized structure

    1 = first pass

and $\bar{\sigma}'$ has its usual meaning.

Figure(8.15) shows the effect of Z on the mean hot rolling plane strain yield stress for samples deformed during the third pass of a 3x25% rolling schedule. The behaviour of $\bar{\sigma}'$ in the third pass is very similar to the one found during the second pass. The third pass samples can be subdivided according to their microstructure entering the pass into two subgroups;

fully unrecrystallized and partially recrystallized samples. These subgroups can be further subdivided according to their previous deformational history. This will be later discussed in more detail.

Similarly to the second pass mean plane strain strength, it can be observed from figure(8.15) that

$$\bar{\sigma}'_{3unrex} \simeq 1.15 \; \bar{\sigma}'_{2unrex} \tag{8.11}$$

where the subscripts stand or

    3unrex = third pass, unrecrystallized structure

    2unrex = second pass, unrecrystallized structure

It is acknowledged that the microstructure entering the third pass may have several different populations of material with different amounts of stored energy. The

spectrum may vary from material unrecrystallized with accumulated strain from the previous two passes to material which was freshly recrystallized after the second pass. However, it can be clearly seen from figure(8.15) that unrecrystallized samples entering the third pass have a higher $\bar{\sigma}'$ than the ones with mixed structure.

Figure(8.16) shows the set of points for first,second and third passes plotted together as a function of the Zener-Hollomon parameter, hence a comparison between the three sets is readly available. Details of the three passes are given in tables (8.1) to (8.3).

The Zener-Hollomon parameter includes the effects of both strain rate and rolling temperature on the strength of the material being hot rolled. The average equivalent strain rate varied from a lower limit of 3.22 s^{-1} to an upper one of $4.75s^{-1}$. Therefore, the temperature factor is much more critical in such rolling schedules than the average equivalent strain rate. Hence, figure(8.16) was re-plotted as shown in figure(8.17) from which the dependence of the mean plane strain stress on the average rolling temperature can be immediately assessed.

8.6 Microstructure of Hot Rolled AISI316L.

Figure(8.18) shows the recrystallization process occurring in a sample hot rolled in a single reduction of 25% in thickness and air cooled to room temperature. Figure(8.18-a) shows that only very few nuclei are present in the deformed structure mostly in the grain boundaries. In figure(8.18-b) a more advanced stage of the process of static recrystallization has taken place. It can be seen that there is a simultaneous presence of originally deformed grains, recrystallizing grains and what appears to be recently created nuclei. This suggests that nucleation process happens at the same time as recrystallizing grains are sweeping away the deformed zone of the specimen. In figure(8.18-c) most of the originally deformed grains have either been eaten away or have decreased their original size considerably.

Figure(8.19) shows the microstructure found in a sample hot rolled in a 3 passes schedule with an inter-pass air cooling period of 15s approximately. Figure(8.19-a) shows an overall view of the structure. It can be seen that several different types of microstructures co-exist in the same sample. This multiplicity of structures can be more readily observed from figure(8.19-b). The structures can be subdivided into at least 4 groups:

a- Structure 1 is the remainder of the original material. It has an accumulated strain equal to 0.99

b- Structure 2 is the material which recrystallized

after the first pass and was deformed in the following

pass. This results in an accumulated strain of 0.66

c- Structure 3 is the one which results from recrystallization after a second pass was given to the material

d- Structure 4 is the part of the sample to recrystallize only after the third pass.

Figure(8.20) shows some metallographic features of a sample deformed in a 2x25% reduction schedule and quenched immediately after the second pass. It can be seen that it is possible to assess the fraction recrystallized that occurred during the resting period between passes 1 and 2. The original grains (structure 1) differ mainly in size from the recrystallized ones (structure 2). As the fraction recrystallized increases, the original grains are eaten away. Therefore the above mentioned size difference tends to disappear. Hence, it becomes more difficult to count the recrystallized volume fraction. The effect is illustrated in figure(8.20-b) in which the experimental recrystallized fraction was found to be 77%.

8.7 Structural Evolution During Hot Rolling.

8.7.1 Structure After the First Pass.

The dependence of the fraction recrystallized on the average rolling pass temperature for samples deformed in a single pass of 25% reduction in thickness is shown in figure(8.21). The samples were held in air for 15s and either quenched or air cooled to room temperature. It can be observed from figure(8.21) that the higher the average rolling temperature the greater the fraction recrystallized. It is also clear that the air cooled material recrystallizes nearly twice as much as the samples cooled in air for 15s only. It is also expected that only a very small portion of the material (less than 5%) will recrystallize if a pass is given at temperatures lower than 1000 C. Figure(8.22) shows the original grain size effect on the recrystallized fraction of samples hot rolled under identical conditions.

The dependence of the average grain size during recrystallization on the average hot rolling temperature is depicted in figure(8.23). Samples 200μm original grain size present a constant average grain size during recrystallization (~130 μm) for temperatures higher than 1100 C. Samples cooled in air to room temperature have a smaller average grain size during recrystallization than the samples cooled in air for 15s and quenched. Specimens with 100 μm original grain size give a constant average grain size of 70 μm (air cooled and quenched) and 50 μm(air cooled to room temperature). The temperature interval above which the average grain size during recrystallization becomes constant is 1060 to 1090 C.

Figure(8.24) shows the evolution of the average grain size during recrystallization, $\bar{d}$, with the sample fraction recrystallized. It can be seen from figure (8.24) that $\bar{d}$ changes very rapidly with the recrystallized fraction. A constant $\bar{d}$ plateau is achieved soon after 50% of material has recrystallized. Samples of original grain size 100 μm tend to group together in a same experimental band regardless of the air cooling period given to them. Full details of the results obtained for single pass samples are given in table (8.4).

8.7.2 Structure after Second and Third Passes.

Figure(8.25) shows the average grain size during recrystallization dependence on the fraction recrystallized after a second pass is given to a sample recrystallized after the first pass. In such a case, the material which is being averaged comprised fully deformed grains with accumulated strains from the first and second passes or from the second pass only, and material which is freshly recrystallized after the second pass. It can be noticed from figure(8.25) that $\bar{d}$ is roughly constant for recrystallized fractions higher than 30%. For fraction recrystallized less than 30%, it seems that there is a sharp increase in $\bar{d}$ as the fraction recrystallized decreases. For most values of fraction recrystallized (X > 30%), $\bar{d}$ seems to be fairly constant and within an experimental band 20 μm wide. Therefore, $\bar{d}$ varies from ≈30μm to 50 μm for most values of

recrystallized fraction, regardless of the specimen mechanical history. The same apparent insensitiveness to previous mechanical treatment is illustrated by figure(8.26) in which the average recrystallizing grain size is plotted against the average hot rolling temperature.

Figure(8.27) shows the dependence of $\bar{d}$ on the material statically recrystallized after the third pass. Figure(8.28) shows the average rolling temperature effect on $\bar{d}$ for samples hot rolled in 3x25% schedule. The overall picture is very similar to the one found for second pass samples. The material seems to behave fairly insensitive to previous mechanical history if the recrystallized fraction is higher than 20%. Detailed accounts of single-pass, two-pass and three-pass schedule samples are given in tables (8.4), (8.5) and (8.6) respectively.

It is therefore expected that after a second pass a sample will have an average grain size during recrystallization equal to $\approx 40\mu m$ whereas $\bar{d}$ after the third pass would decrease to 25 μm.

A typical example of structural behaviour during hot rolling is the one from samples R80,R79 and R81 represented in figures (8.29) and (8.30). Immediately before the third pass is given, the average grain size during recrystallization is $\approx 35\mu m$ whereas it is 25 μm before water quenching. The fraction of material to recrystallize after a given pass seems to be more sensitive to the thermo-mechanical treatment than $\bar{d}$. Figure(8.30) shows that nearly 50% of the material recrystallizes before the second pass. The sample

recrystallizes 50% again before entering the third pass. However, only 12% appears to recrystallize after the third pass is given.

An attempt to further investigate the effect of the fraction recrystallized on the rolling load was made as illustrated in figure(8.31).

Samples were rolled in a two-pass schedule. The first 25% reduction pass was given at several different measured slab centre temperatures at the roll gap entrance, T_{entry}^{mc}. The second pass was a 21% reduction one given at T_{entry}^{mc} constant (950 C). Samples were immediately quenched after the second pass for metallographic studies.

Figure(8.31) shows that the second pass rolling load does not vary linearly with the fraction of material recrystallized before the second pass. It is apparent that the rolling load becomes approximately constant at recrystallized fraction values higher than 50%. However, a rapid increase in the rolling loads is seen to occur for recrystallized fraction less than 50%. Figure(8.32) shows that a sigmoidal type curve is obtained when the rolling loads measured in the second pass are plotted against T_{entry}^{mc} for the first pass. It is again apparent that no recrystallization occurs when the samples are deformed at temperatures lower than 1000 C.

8.8 Hot Rolling Simulation.

8.8.1 Introduction.

A modified version of Leduc's computer programme (Leduc,1980) was used for simulating the hot rolling strength and structural evolution of the experimental data available in the present work. The modifications made in the original programme were mainly due to the use of AISI316 as the material for simulation. Details of the alterations carried out are given in appendix (8.2). This section only presents the results from the simulation performed using the modified version of the programme.

8.8.2 Strength During Hot Rolling.

The computed mean plane strain strength dependence on the measured one is shown in figure(8.33). Samples were deformed in schedules involving 25% reduction in each pass as comprehensively described in table (8.4). It can be observed from figure(8.33) that the points plotted are evenly scattered around a 45 degrees straight line. The correlation factor was found to be 0.92.

It is convenient to have the average pass temperature plotted against the measured temperature in the centre of the sample at the roll gap entrance as shown in figure(8.34). The figure can be used as a means of obtaining

a quick assessment of the average pass temperature from the centre measured ones. The experimental points from figure(8.34) can be reasonably fitted by

$$T = 104 + 0.87 \; T^{mc}_{entry} \tag{8.12}$$

The correlation coefficient for the data examined and plotted in figures(8.34) is 0.994. The differences between average and measured centre temperature decreases as the rolling temperature decreases. At temperatures lower than 900 C, it can be said that the average temperature is approximately equal to T^{mc}_{entry}.

8.8.3 Structural Evolution.

Figure(8.35) displays the correlation existent between computed average grain size and the measured average grain size during recrystallization. Samples were deformed mostly in a 25% single pass schedule with essentially two different original microstructures; 100 and 200 µm. It can be observed from figure(8.35) that there is an overall agreement between predicted and measured values irrespective of differences in the original structure. Figure(8.36) exposes the same comparison as for figure(8.35) but for samples deformed in a double pass schedule. Most of the samples tested had 100 µm original grain size. The agreement between measured and predicted average grain sizes during static recrystallization seems to be reasonable for the experimental material available.

Figure(8.37) shows the correlation between predicted and measured average grain size during recrystallization for samples deformed in a 3x25% pass hot rolling schedule. Once again, the agreement between predicted and experimental values is reasonable. Samples tended to present a similar microstructure after the third pass regardless of the prior thermo-mechanical treatment given to them. The microstructure could be further refined if the samples were allowed to recrystallize in significant amounts (>50%) after the third pass. However, third pass rolling temperatures were well below 1000 C. Hence, static recrystallization could not proceed towards values of 50% or higher.

The computed fraction recrystallized after passes number 1, 2 and 3 is plotted against the measured fraction recrystallized as shown in figures(8.38), (8.39) and (8.40) respectively. Figure(8.38) presents the largest variety of fraction recrystallized values. As the number of passes increases, the recrystallized fraction values tend to fall within a relatively narrower band. This is most noticeable for the three-pass samples. The general agreement between predicted and measured values of recrystallized fractions is reasonable for all the three data set existent.

The efficiency of the heat flow model can be estimated as indicated in figures(8.41), (8.42) and (8.43). The computed and measured centre temperature at the roll gap entrance should , ideally, coincide. In figures (8.41) to (8.43), the experimental inter-pass rolling times were used to determine

T_{entry}^{mc}. Therefore, the figures can be used to visualize how well the heat flow model is simulating the temperatures found during the hot rolling process. Tables (8.4) to (8.6) give details of all samples and rolling schedules used for experimental data acquisition in this section.

8.9 Discussion.

8.9.1 Introduction.

8.9.1.1 Original Grain Size.

The grain size of the 316L steel showed a constant value (d_o~100µm) when reheated for 1/2 hour at temperatures ranging from 950 to ~1200 C. However, in the relatively small temperature interval of 100 C (from 1200 to ~1300 C), the grain size increased from 100 to ~300µm (figure(8.1)). Similar behaviour had been observed in heat treatment of samples for plane strain compression tests (Colás,1983).

The sharp increase in d_o in a limited range of annealing temperatures could have been caused by the pinning effect of fine precipitates. This has been reported to be the case for Ti-steels (Leduc,1980; Leduc and Sellars,1981). It has also been shown that grain coarsening temperature in austenite is

controlled by the stability of precipitates which are
pinning the grain boundaries (Weiss et al.,1981). On the
other hand, Cole(1979) and Al-Jouni(1983) reported a
coarsening temperature for austenitic stainless steels of
~1150 and ~1070 C respectively. However, no comments were
made on the pinning of grain boundaries as the cause for
this relatively high coarsening temperature.

Thin foil samples of as-received material of 316L were
analysed and evidence was found of the presence of TiN
precipitates (approximately 0.1μm in size) (Castro,1983).
The presence of $M_{23}C_6$ precipitates 0.05 μm in size was also
detected (Castro,1983). These could have led to grain
boundary pinning. It was therefore thought that this
mechanism could be used as an explanation for the annealing
behaviour shown in figure(8.1).

8.9.1.2 <u>Surface Heat Transfer Coefficient During Air
Cooling.</u>

The heat transfer coefficient, (H_s), at the slab
surface needs to be determined as a function of the slab
surface temperature so that cooling behaviour during rolling
may be simulated. Harding(1976) carried out measurements of
H_s on 304 austenitic stainless steel for an 'infinite' slab
with insulated ends and for a short slab (~5" long). He
arrived at an analytical expression which describes H_s in
terms of the sample surface temperature. His equation was
later used for simulation of air cooling in C-Mn steels

(Leduc,1980). Apparently, there was no need for altering either the radiation or the convective coefficients of Harding's equation in order to apply it to the case of C-Mn steels. On the other hand, Harding himself had previously pointed out that H_s for the short 304 slab was consistently higher than the one obtained for the 'infinite' slab. He attributed the differences in H_s to the effect of an axial heat flow in the short slab, which his two-dimensional heat flow model could not have accounted for. Therefore, an apparent H_s was obtained from measurements in the short slab.

Harding's equation for H_s, derived for infinite slabs, could not be used for 316L steels. The two steels cool differently in air (figure(8.2)). This was thought to be caused by an axial heat flow (316L slabs were ~4" long). It is also known that 316L steels form a thinner oxide layer than 304 steels during heating. The heat flow programme was run to simulate air cooling of a 316L slab 4"x10" cross section. In this way, it was possible to assess the time needed for the thermal gradient to penetrate 2" unidirectionally. The time needed for penetration was found to be ~40s. However, differences in air cooling behaviour between 316 and 304 steels started at ~25s. This suggests that a thinner oxide layer thickness causes 316L to cool faster than 304 steel at times between 25 and 40s. After 40s, the axial heat flow adds to the effect of the oxide layer thickness effect and the result is an even faster cooling rate for the 316L sample. Clearly, the radiation coefficient may not be

affected as much as the convective one. In fact, the measured emissivity for 316L (0.81) is in good agreement with the reported value for 304 steels (0.84) (Harding,1976). However, measurements of the convective coefficient may have produced an over-estimated value. The use of equation (8.1) for simulation of air cooling produced results in agreement with the experimental data available (figure(8.5)).

8.9.2 Hot Rolling of 316L.

8.9.2.1 Strength During Rolling.

The mean plane strain strength during hot rolling, $\bar{\sigma}'$, remained constant for samples with original grain sizes varying from ~260 to ~60µm. For original grain sizes less than 60µm, there is an increase in $\bar{\sigma}'$ as shown in figure(8.10). The experimental data can equally be fitted by a smooth curve or by two segments of straight lines. At first $\bar{\sigma}' \alpha d_o^{-0.05}$ and when d_o is smaller than ~60µm $\bar{\sigma}' \alpha d_o^{-0.15}$. There is no previous report in the literature concerning the effect of the original grain size on the mean plane strain strength during hot rolling. On the other hand, it has been seen that the effect of the original grain size on σ_p is not completely clear (section (2.2)). The present results from hot rolling tests are compared with the ones obtained from several authors testing in torsion, axisymmetric compression

and plane strain compression (figure(8.45)). It is interesting to note that for original grain sizes bigger than ~40μm, the σ_p values change only very slightly. However, there is a contradictory picture for d_o less than ~40μm. Colàs testing 316L steel in plane strain compression reports an increase in σ_p compatible with a -0.15 d_o exponent whereas Foster(1981) testing Nb-steels at 900 C, has found no change at all of the original -0.05 d_o dependency. The coincidence in the 316L behaviour suggests that a common factor is influencing both available sets of results. In favour of this was the fact that in both cases (present work and Colàs samples) the specimens were heat treated similarly. The thermo-mechanical treatment consisted of pre-heating for 1/2 hour at ~1180 C and rolling to the final thickness of ~10mm. The samples were then quenched and reheated for 15 minutes at 970 C to give a fine starting microstructure before re-rolling. Thin foil examination of the hot rolled and quenched sample revealed the presence of $M_{23}C_6$ precipitates in the matrix and in the grain boundaries (Castro,1983). The precipitates may not have been completely taken into solution in the heat treatment given at 970 C (Deighton,1970) which could explain the increase in the strength at low value of d_o.

The rolling load varied linearly with the reduction in passes given at a measured entry temperature of ~950 C (figure(8.11)). On the other hand, the mean plane strain strength during rolling reflected the stress-strain curve of AISI316L tested at 950 C (figure(8.12)). In fact,

at $\epsilon=0.2$, an isothermal test gives a lower value of $\bar{\sigma}'$ ($\sim$210MPa) than the one found for hot rolling samples ($\sim$218MPa). The same behaviour is consistently repeated at $\epsilon=0.4$ (hot rolling $\bar{\sigma}'$ 240MPa as opposed to $\sim$230 from plane strain compression). Clearly, the average rolling temperature is lower than the one for plane strain compression test. As seen previously, a difference in temperature of $\sim$20 C accounts for the discrepancies in $\bar{\sigma}'$ values of $\sim$10MPa. Consequently, the average hot rolling temperature can be estimated as $\sim$930 C. Estimation using figure(8.34) gives $T\sim$931 C.

The mean plane strain strength for a single pass of 25% reduction varies linearly with the logarithm of the Zener-Hollomon parameter (figure(8.13)). The strain rate for each pass was kept approximately constant (3 to $5s^{-1}$). Therefore, diagram (8.13) really represents the effect of the average rolling temperature on $\bar{\sigma}'$. The measured $\bar{\sigma}'$ values are in agreement with the ones obtained from isothermal plane strain compression samples. Additionally, σ_p from isothermal plane strain compression samples closely follows $\bar{\sigma}'$ obtained from rolling samples ($\sigma_p \approx 165$ at $Z=10^{18}$ and $\sigma_p \approx 270$MPa at $Z=10^{21}$).

Figure(8.13) also shows that $\bar{\sigma}'$ is not affected by the original geometry of the samples which varied in thicknesses from 25mm to 14mm. The original structure was kept constant before the first pass (original grain size 100μm).

The structure entering the second pass is dependent on the first pass average temperature, since the time interval

between passes was kept constant at ~15s. If the average temperature of the first pass is below ~1000 C, then it is likely that there will be no significant amount of recrystallization between first and second passes (see figure(8.21)). The unrecrystallized material will accumulate the strain given in the two passes and the mean plane strain strength will increase considerably in the second pass. If an isothermal rolling schedule could be carried out experimentally at $Z \sim 10^{22}$ (~830 C), $\bar{\sigma}'_1$, mean plane strain strength for the first pass would be 270 Mpa and $\bar{\sigma}'_2$ for the second pass would be ~320MPa (figure(8.16)). These results are confirmed by isothermal plane strain compression tests performed at $Z \sim 10^{22}$. The result from plane strain compression gives $\bar{\sigma}'_1 \sim 270$MPa and $\bar{\sigma}'_2 \sim 330$MPa respectively which are in agreement with the values obtained for hot rolling. Structures partially recrystallized will produce an intermediate $\bar{\sigma}'_2$. This can vary between the lines for the first and the second passes. Attempts to correlate $\bar{\sigma}'_2$ to the fraction of material recrystallized entering the second pass showed considerable scatter with apparently no systematic trend.

The same line of reasoning is applicable to samples which were given three passes. The partially recrystallized samples gave $\bar{\sigma}'_3$ always higher than $\bar{\sigma}'_2$ for the same value of Z. Coincidently, the fraction recrystallized entering the third pass rarely exceeded 40%. If a sample entering the third pass had 100% fraction recrystallized, then its $\bar{\sigma}'_3$ should give a value coincident with $\bar{\sigma}'_1$, as shown when the

14mm samples were rolled (figure(8.13)).

If an isothermal 3 pass hot rolling schedule is carried out at $Z \sim 10^{23}$ ($T \sim 750$ C at $\dot{\varepsilon} \sim 5$ s^{-1}), the mean plane strain strengths in each pass are the ones given in the table below.

Table

Pass Number	Mean Plane Strain Strength, MPa		
	Rolling (Experimental)	Plane Strain(*) (Theoretical)	Axisymmetric(**) (Theoretical)
1	309	360	318
2	355	392	386
3	400	404	410

* extrapolated using equations (2.5) to (2.11); (6.1) to (6.4) and (6.23).

** calculated using σ_p and $\sigma_{0.1}$ dependence on Z from figures (9.4) and (9.5).

The table also includes values from theoretical predictions for isothermal plane strain and axisymmetric compression tests. The mean plane strain strength during hot rolling can be simulated by the predictions from the axisymmetric compression tests. However, it is important to note that the difference $(\bar{\sigma}'_2 - \bar{\sigma}'_1)$ for rolling (46 Mpa) can be simulated by both kinds of tests; i.e. plane strain (32 MPa) and axisymmetric (68 MPa).

On the other hand, $(\bar{\sigma}'_3 - \bar{\sigma}'_2)$ for rolling (45 MPa) is

considerably higher than the ones obtained for plane strain (12 MPa) and for axisymmetric compression (24 Mpa). This suggests that another strengthening component is acting on the third pass in order to increase $\bar{\sigma}'$. At this stage, it is thought that $M_{23}C_6$ precipitation strengthening may be occurring (Castro,1983). The relative importance of work hardening and other strengthening components in the value of the mean plane strain strength during rolling is currently being further studied in the Department (Dutta, 1983).

8.9.2.2 Microstructural Evolution During Hot Rolling.

Diagram (8.46) shows schematically a microstructural evolution during a 3 pass hot rolling schedule. The material deformed in the first pass will recrystallize statically in the interpass period giving rise to two new structures. The recrystallized structure entering pass 2 (structure 1) will follow a similar path to the original material. After pass 2, it will split into 2 new structures (1.1 and 1.2) (see diagram (8.46)). On the other hand, structure 2 will finish giving a fully recrystallized population (structure 2.1) and a deformed one with the accumulated strain from the previous 2 passes. Therefore, there will be 4 different kinds of structure entering pass number 3. However, the fully recrystallized material will only be distinguishable by its recrystallized grain size. Consequently, it may be mixed and observed on the microscope

as a single structure with a grain size averaging both recrystallized populations. The same is likely to happen with the deformed structures. Deformed grains are continuously being eaten away by the recrystallizing ones. Therefore differences between deformed populations may not be readily apparent.

After pass number 3 one would expect 6 different types of structures (from structure 1.1.1 to 2.2.2). However, for the same reasons mentioned above, it is likely that only 4, or perhaps even only 3 structures will be noticeable under the optical microscope.

Assume that the first pass is given at an average temperature of 1130 C. The t_{50} can be estimated as ~7s from figure(6.18). The material, however, cools continuously during the interpass time. The second pass average temperature is assumed to be ~1025C (see diagram (8.46)). Therefore, it is reasonable to use an average interpass temperature of ~1080 C to estimate the static transformations occurring in that period. The equivalent t_{50} at 1080 C is ~10s. Therefore, a 15s air cooling period yields X_1~65%. The recrystallized grain size can be estimated from figure(6.24) as ~75μm. Consequently, the slab will be entering the second pass with ~65% of material with a d_{rex}~75μm (population 1) and ~35% with an accumulated strain of 0.33 (population 2). These predictions agree with values reported in figure(8.21). According to this diagram, for T~1130 C, the fraction of recrystallizedmaterial to enter the second pass is ~75%. Clearly, there are no

difficulties in assessing X_1 metallographically, although differences between d_{rex} and the original grain size are not considerable. Therefore, the microstructural evolution after the first pass can be used for checking the modelling performance.

A similar line of reasoning could be applied to the structure recrystallizing after the second pass. Accordingly, structure 1 will produce 12% recrystallized material with 45μm grain size (structure 1.1) and a deformed material (structure 1.2) with 75μm grain size. On the other hand, structure 2 would consist of some material metadynamically recrystallized, producing 16% with 30μm grain size (structure 2.1) and a double deformed material (9%). Clearly, structures 1.1 and 2.1 would not be easily distinguishable from each other when examined on the optical microscope. Therefore, the net effect would be the observation of 3 microstructures. The recrystallized structure having a grain size in between 45 and 30 μm. However, the differences in grain sizes between recrystallized structures (~40μm) and the one with 0.33 strain (~75μm) is still clear. After the third pass, the recrystallized structure which has been deformed in the second pass will be ~30μm in size. At the same time, there will be recrystallized material of similar grain sizes (30 to 40μm). Therefore, there will be possibly an over-estimation of the fraction recrystallized measured in these circumstances. The present effect is clearly illustrated in figure(8.19b) which shows that grains number 3 are hardly

different from grains number 4.

A further attempt to connect the microstructure entering a pass with the rolling load was performed as seen in figure(8.31) and (8.32). The samples were pre-rolled in a first pass at several temperatures and cooled in air. Then a second pass was given at ~950 C. The samples had the same original geometry and grain size. The draft was kept constant, thus the rolling load was directly proportional to the mean plane strain strength. The samples were quenched immediately after the second pass. The fraction of material to recrystallize between passes 1 and 2 could be then measured, since they differ considerably in size (figure(8.20)). Figures(8.31) and (8.32) show that as the fraction recrystallized increases, the load in the second pass decreases from ~13 to 10 Tonnes (a factor of ~1.3). The same proportion of change is observed in the mean plane strain strength (figure(8.16)) at $Z \sim 10^{20}$ (T~950 C), i.e., from 280MPa to 230MPa, a factor of ~1.22. It has been observed that the decrease in load is not linearly related to the increase in the recrystallized fraction. This is in agreement with results from double deformation torsion tests (Barraclough and Sellars,1979).

8.9.3 Hot Rolling Simulation.

8.9.3.1 Simulation of Microstructural Evolution.

The equations used to modify the original C-Mn version of Leduc's programme (Leduc,1980) are concerned with:

a- time for 50% recrystallization

b- recrystallized grain size

c- isothermal grain growth

The set of equations initially derived from the plane strain compression tests for the t_{50} were given by expressions (6.8), (6.9) and (6.12). The apparent activation energy was found to be $475\pm29KJ/Mol$, which is similar to the one reported by Towle and Gladman(1979) (525KJ/Mol). More recently, a lower value of activation energy for recrystallization of 316L steel has been reported (Ryan et al.,1982). However, their value (336KJ/Mol) seems to be extremely low, even when compared with Q_{rex} for 304 stainless steels ($\sim$410KJ/Mol). It was then decided to use an average activation energy for recrystallization equal to 500KJ/Mol in equation (6.8). Accordingly, expression(6.12), which describes the critical strain, was slightly altered so that $Q\sim$134KJ/Mol.

It was initially found that the __apparent__ activation energy changed with the prior strain applied to the sample (figure(6.20). Later experimental work was carried out to measure the actual activation energy at strain lower and

higher than the critical one, ϵ_*. In fact, Q_{rex} was confirmed to be independent of the prior strain applied to the sample, and its value is ~500KJ/Mol. This coincided with the apparent activation energy used in equation (6.18) However, a constant activation energy for recrystallization implies that the t_{50} dependence on Z changes with the strain applied to the sample. In the present work, it seems plausible to use $Z^{-0.6}$ to correlate the limited data available on t_{50} for $\epsilon > \epsilon_*$. It is interesting to note that the same Z exponent has been used for C-Mn steels (Sellars,1980).

The overall results indicate that the new set of equations for t_{50} simulation could be:

$$t_{50} = AZ^{-0.38}d_o^{1.3}\epsilon^{-3.6} \exp\frac{500000}{R\ T} \qquad \epsilon < \epsilon_* \qquad (8.14)$$

and

$$A = 4.00 \times 10^{-15}s^{0.62}\mu m^{-1.3}$$

$$t_{50} = B\ Z^{-0.6} \exp\frac{500000}{R\ T} \qquad \epsilon > \epsilon_* \qquad (8.15)$$

where

$$B = 1.6 \times 10^{-8}s^{0.4}$$

and also

$$\epsilon_* = 0.01\ Z^{0.06}\ d_o^{0.36} \qquad (8.16)$$

Either set of equations ((8.14) to (8.16) or (6.8) to (6.12)) can be used for simulation of hot rolling of 316L.

The differences in t_{50} predictions are of marginal significance as demonstrated by the table below. However, since it has been found that the activation energy for recrystallization seems to be independent of the prior strain applied to the sample, equations (8.14) to (8.16) should be preferred to the previous set.

Table

Temperature, C	t_{50},s*		
	eq.(8.15)	eq.(6.9)	experimental
1100	2.0	3.0	2.0
1025	6.3	8.1	7.5
950	22.5	23.9	19.0

* Strain rate used to calculate $t_{50} = 5s^{-1}$

The isothermal grain growth of austenitic stainless steel has been previously reported to be reasonably described by a lower than 10 grain size exponent (Martin and Holmes,1979; Towle and Gladman,1979). In the present work, the limited experimental data available on grain growth at the time, could be described by

$$d^3 = d^3_{rex} + (5.25 \times 10^{14}) \ t \ exp\frac{-305000}{R \ T} \qquad (8.17)$$

Therefore, equation (8.17) was used in the programme for hot rolling simulation. Clearly, no physical significance is claimed for the d_{rex} cubic exponent. In fact, a later review showed that equation(3.29) also described isothermal grain

growth of 316L steels reasonably well. Coincidentally, the activation energy for grain growth obtained from equation(3.29) (~320KJ/Mol) is of the same order of the one used in equation(8.17). Figure(8.47) compares the performance of the two equations. It can be seen, the discrepancies between both equations are of the order of 1μm at 1100 C after 25s annealing. Clearly, these differences cannot be considered significant. In addition, hot rolling schedules were mostly carried out with time intervals of ~15s. Nonetheless, equation(3.29) is fundamentally better than equation(8.17) and should be used in any further work of simulation.

The modifications introduced in the original C-Mn programme are shown in appendix(8.2). It can be seen from the programme listing that k, the Avrami equation time exponent, equal to 1 was used in the simulation. The model assumes that the strain applied during rolling is homogeneously distributed on the cross section of the sample. In addition, it has been previously shown (figure(7.2)) that k=1 on iso-strain lines. Therefore, it was felt justifiable to use k=1 in the modelling of 316L hot rolling.

Another point to be mentioned is that it has been shown that the kinetics of static recrystallization of a sample non-homogeneously deformed cannot be described by an Avrami type of equation (figure(7.14)). However, as strain is assumed to be uniform during the hot rolling simulation process, an Avrami equation was used in the present programme.

Despite the simplifications mentioned above, theoretical

values of fraction recrystallized and average grain sizes during recrystallization are in reasonable agreement with the experimental ones (figures(8.35) to (8.40)). This is particularly true for the single pass samples (figure(8.35) and (8.38)) which suggests that the equations used for microstructural evolution are at least of the correct form.

8.9.3.2 Strength and Temperature Simulation.

The expressions used in the computer programme for simulation of hot rolling loads are given by equations(6.1) to (6.4) together with equation(6.23). The necessary modifications carried out to the original version of the C-Mn programme (Leduc,1980) are detailed in appendix(8.2). The agreement between theoretical predictions and measured hot rolling loads is shown in figure(8.33) which summarizes all the results available for laboratory hot rolling data. Clearly, the structure entering each pass could consist of fully deformed, fully recrystallized or a mixed microstructure. Nonetheless, the largest discrepancy between predicted and experimental values is of the order of ~15%. However, most points are within a 10% scatter band. This is in agreement with the simulation of rolling loads carried out for C-Mn steels (Leduc,1980). However, C-Mn steels often recrystallize completely in between passes. Puchi(1983) working with aluminium produced a somewhat larger discrepancy (~20%) between predicted and experimental results for industrial hot rolling loads. This was however

attributed to a non-homogeneous strain distribution effect, since his industrial simulation work involved the deformation of thick plates in the roughing passes. Comparison between several theories for hot rolling calculations shows that the disagreement between them varies from a minimum of ~10% to a maximum of ~15% (Larke,1957). The difference varies as a function of the ratio of roll radius: final thickness, increasing as the ratio increases.

The air cooling interpass rest period can be reasonably simulated using equation(8.1). Figures(8.41) to (8.43) show the agreement between measured centre temperature at the roll gap entry and the theoretical ones. In obtaining these results, it was necessary to match the measured and the theoretical centre temperature at the first pass for some samples. This was, however, expected since the precise time for starting the rolling schedule is not exactly clear. The thermocouple inserted in the middle of the sample does not detect instantly when the sample is being removed from the pre-heating furnace. In several samples, the procedure was taken of marking the temperature chart record at the moment of the withdrawal of the samples. In these cases, theoretical and experimental temperature measured at the center of the slab cross section entering the roll gap were in good agreement. Clearly, a marking device must be used in future simulation work, possibly by attaching a second thermocouple to the pair of tongs.

8.10 Conclusions.

1- Austenitic stainless steel 316L presented a coarsening temperature of ~1200 C. This is supported by recent investigation work on austenitic stainless steels (Colàs,1983; Al-Jouni,1983; Cole,1979). It is thought that grain boundary pinning by precipitates is responsible for the observed behaviour.

2- The expression

$$H_{theor} = -0.0368 \ (T_s - R_t) - -0.4579 \times 10^{-10} \ (T_s^* - R_t^*)$$

seems to describe suitably the surface heat transfer coefficient for the slabs used in the present investigation. The use of this equation in the two-dimensional heat flow model leads to the accurate prediction of the air cooling behaviour.

3- The mean hot rolling plane strain strength is solely dependent on the Zener-Hollomon parameter and the original grain size. An exponential type of equation can be used to describe most experimental data available in the present work.

4- The hot rolling load seems to be significantly dependent on the fraction of recrystallized material entering a given pass. The rolling load decreases as the fraction recrystallized increases.

5- Partial recrystallized structures from laboratory hot

rolling schedules seem to be reasonably simulated by Leduc's programme, provided the correct set of equations for t_{50}, d_{rex} and grain growth are used. Likewise, hot rolling loads and temperature evolution can be satisfactorily simulated.

CHAPTER 9

Hot Rolling Simulation by Plane Strain Compression Tests

9.1 Introduction.

In the present chapter, results on the simulation of hot rolling microstructural evolution from plane strain compression tests are presented. The chapter itself may be subdivided into three sub-sections concerned with:

a- comparison between plane strain and axisymmetric flow stress curves

b- preliminary work needed for carrying out plane strain compression simulation. This section refers mainly to heat transfer coefficient determination and creation of furnace conditions needed for simulation of the cooling curve during hot rolling.

c- hot rolling simulation by using plane strain

compression tests to investigate microstructural changes.

9.2 Plane Strain versus Axisymmetric Compression Tests.

The equations used for simulating stress-strain curves for AISI316L were derived from plane strain compression test results. At that time, the insensitiveness of the strain to the peak stress, ε_p, on the Zener-Hollomon parameter, Z, was noticed. It was thought that the strain distribution in a sample tested under plane strain compression could have been the cause of the above mentioned ε_p behaviour. Therefore, a few axisymmetric tests were conducted for comparison purposes with the plane strain compression ones.

Figure(9.1) shows the stress-strain curves for samples tested under axisymmetric compression. The temperatures used during testing coincided with the ones employed earlier for plane strain compression. Most curves finish at an equivalent strain of 1.0 or marginally higher strains. The strain to peak stress is not always clearly defined since the curves become horizontal for a wide strain band. After this, the stress gently decreases as the strain increases further. Hence, it was arbitrarily assumed that ε_p is the minimum value of equivalent strain to give the maximum stress, σ_p. The ε_p values measured in the manner defined above are indicated in diagram(9.1) from which it

can be clearly seen that ε_p is dependent on the Z value. A comparison between the dependence of ε_p on Z from axisymmetric samples with the ones for plane strain compression samples is shown in figure(9.2). Axisymmetric ε_p values are always bigger than plane strain compression ones and

$$\varepsilon_p \quad \alpha \quad Z^{0.15} \tag{9.1}$$

rather than being nearly constant as for plane strain compression tests.

However, plane strain and axisymmetric compression maximum stress are coincident for a given value of Z. This is shown in figure(9.3) for a particular Z, whereas figure(9.4) shows the comparison between σ_p for plane strain with the ones obtained for axisymmetric test for a broader range of Z. It clearly follows that $\sigma_a < \sigma_{ps}$ when $\varepsilon < \varepsilon_p$ where the subscripts a and ps stand for axisymmetric and plane strain compression tests respectively. This effect is illustrated in figure(9.5).

9.3 Temperature Evolution During Plane Strain Compression.

In order to simulate hot rolling using plane strain compression testing, it is necessary to know how the material cools both in air and inside the test furnace. This is done by using Foster's two-dimensional finite-difference heat flow model (Foster,1981). The programme provides means

of calculating the average plane strain pass temperatures. A few modifications had to be introduced to the original version due mainly to the differences in thermal properties for the material used in this research. Therefore, new heat transfer equations had to be determined for air cooling and furnace cooling.

Figure(9.6) shows the dependence of the surface heat transfer coefficient (H_s) on the sample surface temperature (T_s) for a specimen held in air in the Servo-test machine carriage arms. The full line represents the equation which best fitted the experimental points

$$H_s = -0.0426 (T_s - R_t) - 0.3187 \times 10^{-10} (T_s^* - R_t^*) \qquad (9.2)$$

where R_t is the room temperature, assumed to be 293K. The above expression was used in the heat flow computer programme for air cooling simulation. Figure(9.7) shows an air cooling curve for a sample pre-heated to 1170C. It was found that a slightly modified convection coefficient gave a better agreement for times longer than 80s. Therefore, the equation finally adopted for simulation purposes was

$$H_s = -0.045 (T_s - R_t) - 0.3187 \times 10^{-10} (T_s^* - R_t^*) \qquad (9.3)$$

The method used for H_s determination was again applied to the situation where the sample was kept inside the furnace for a 'furnace cooling' process. The sample was pre-heated at 1170C, left 5s cooling in air and transported to the test furnace. There, it was held at two different temperatures; 848 and 228C. An expression similar to equation(9.3) was

determined

$$H_s = -0.0644 \ (\ T_s - R_t \ ) - 0.49329 \times 10^{-10} \ (\ T_s^* - R_t^* \ ) \qquad (9.4)$$

and used in Foster's programme(Foster,1981) for 'furnace cooling' simulation. This is shown in figure (9.8) where theoretical and experimental data are compared.

The test furnace temperature chosen for simulation of air cooling during experimental hot rolling simulation was 230C. As a standard procedure, samples were always taken out of the pre-heating furnace, allowed 5s air cooling and transported to the test furnace. Times shorter then 5s are difficult to reproduce. Samples were always deformed in passes of 25% reduction each with 15s 'furnace cooling' periods in between passes. Figure(9.9) shows a typical plane strain compression schedule used in the experimental programme. The full line represents the experimental result whereas the crosses give the values of temperature at the centre of the specimen theoretically obtained. The average temperature for a pass i was calculated as

$$\bar{T}_i = \frac{\bar{T}_{entry} + \bar{T}_{exit}}{2} \qquad (9.5)$$

where $\bar{T}$ means average temperature whilst the sub-scripts entry and exit mean at the start and finish of a plane strain compression pass.

An estimated mean temperature during a test was calculated for guidance purposes only. The expression used was

$$\overline{T} = \frac{T_i + T_{i+1}}{2} \tag{9.6}$$

where the symbols are as represented in figure(9.9). Therefore, after testing and computing the average pass temperature by using equation(9.6), T_i would be compared with $\overline{T}$. This comparison is shown in figure(9.10), which shows that the estimation method above described gives reasonable $\overline{T}$ values for most cases examined.

9.4 Hot Rolling Microstructure Simulation.

It is important to have an appropriate sample test furnace cooling rate which coincides with the one obtained for a hot rolling slab being cooled in air. In this manner, an inter-pass time in the hot rolling schedule will be compatible with the one from plane strain compression tests. As mentioned above, the test furnace temperature for hot rolling air cooling simulation is 230C for AISI316L. Both cooling curves for hot rolling and for plane strain compression samples are plotted together as previously seen in figure (5.13).

Once the cooling rate is correctly set, it still remains to ensure that the original microstructure and strain applied to the sample are the same as for hot rolling slabs.

The microstructural evolution during plane strain compression was followed as described below:

a- samples with 100 and 200 μm original grain size were

given a single 25% reduction pass at temperatures ranging from ~1200C to 1000C. Figure(9.11) shows some of the microstructural characteristics after the first pass.

b- after deformation, samples were held in the test furnace for 15s.

c- finally, the specimens were pulled out of the test furnace, quenched and sectioned for microstructure observation. This consisted of measuring the fraction recrystallized, X, and the average grain size during recrystallization, $\bar{d}$.

In this way, the structure entering the second pass was frozen and quantified. Figure(9.12) shows the dependence of the fraction recrystallized entering pass number two on the temperature for the first pass. This was calculated using equation(9.5). Hence, X can be predicted from diagram(9.12) for a given average test pass temperature, $\bar{T}$. It is clearly seen from figure(9.12) that the material entering the second pass consists of a mixed microstructure, i.e., it has populations of recrystallized grains and deformed grains co-existing in a single microstructure. The recrystallized fraction values vary from ~80% at 1200C to ~5% at 1030C for a material 100µm original grain size. Recrystallization may be considered negligible (< 5%) for $\bar{T}$ < 1030C.

The dependence of $\bar{d}$ on fraction recrystallized is shown in figure(9.13). It clearly illustrates that $\bar{d}$ decreases very rapidly during the early stages of recrystallization (X<20%). A level value of $\bar{d}$ is reached at X ~40 to 50%

regardless of the original grain size.

Figure (9.14) shows $\bar{d}$ as a function of the first pass average temperature. Values of $\bar{d}$ can be read from figure(9.14) for any particular value of $\bar{T}$. Diagram(9.14) together with (9.12) can be used to generate interpolated $\bar{d}$ and X values which in turn may be compared with the ones determined experimentally. In this way, hot rolling microstructural evolution can be simulated. Table(9.1) shows details of comparison between measured $\bar{d}$ and X values with the ones interpolated from the diagrams mentioned above. A graphical comparison is also provided by figures(9.15) and (9.16). In figure(9.15) the measured fraction recrystallized from hot rolling samples ($X_{rolling}$) is compared with the predicted ones from plane strain compression samples ($X_{plane-strain}$). Points are evenly distributed around the 1:1 slope line, meaning that the agreement between the two microstructures is reasonably good. A similar comparison was carried out for $\bar{d}$ as shown in figure(9.16). There again, it seems that $\bar{d}$ values obtained from plane strain compression tests correlate well with the rolling ones.

Figure(9.17) shows the mean plane strain strength during the first pass, $\bar{\sigma}'$, as a function of the Zener-Hollomon parameter. The full line represents equation(8.9) obtained from hot rolling. At the same time, points from isothermal and non-isothermal plane strain compression tests are plotted for comparison purposes. The agreement between $\bar{\sigma}'$ value from isothermal tests and the ones from hot rolling samples is reasonable for average test temperatures ranging

from 1200 to 800C. The same is found to be the case for non-isothermal samples tested at average test temperatures higher than ~1050C.

Figure(9.18) shows the microstructural behaviour of a plane strain compression sample after the second pass. Figure(9.19) shows the dependence of the recrystallized fraction entering the third pass on the average temperature of the second pass. Samples were given 2x25% reduction passes with the usual 15s cooling rest period in between them. After the second pass, the sample was kept inside the test furnace for further 15s and then quenched. The microstructure obtained is expected to represent the material entering a third pass. After quenching, samples were sectioned and the fraction recrystallized and the average grain size during recrystallization were measured.

Figure(9.19) shows that X, the fraction recrystallized, varied from 60% to 15% when the second pass was given at an average test temperature in the range of ~1130 to 980C respectively. The diagram may be used to obtain interpolated X values for any given $\bar{T}$ value.

The average grain size during recrystallization depends on the average test temperature as shown in figure(9.20). There is an increase in $\bar{d}$ values as $\bar{T}$ decreases further from 1050C

because the fraction recrystallized increases. However, $\bar{d}$ remains fairly constant for $\bar{T}$ ranging from 1150C to 1050C. This relative insensitiveness to average second pass temperature contrasts with the fraction recrystallized behaviour. As for figure(9.19), $\bar{d}$ can be interpolated for any $\bar{T}$. This has been done in order to compare X and $\bar{d}$ from hot rolling samples with the ones obtained from plane strain compression samples.

Figure(9.21) shows the comparison of fraction recrystallized for samples hot rolled in 2x25% passes with values X obtained from plane strain compression tests. Details are given in table(9.2). It is clearly seen that most points are evenly scattered around the line of slope 1. However, a few samples produced a X rolling value of ⁻45% whereas the corresponding plane strain compression ones are predicting ⁻30%.

The comparison between $\bar{d}$ measured from hot rolled samples and $\bar{d}$ interpolated from figure(9.20) is shown in figure(9.22). Details of the measured and interpolated $\bar{d}$ values are fully documented for in table(9.2). The range of $\bar{d}$ studied was limited (⁻30 to ⁻60 μm). This is mainly due to the fact that $\bar{d}$ is fairly insensitive to the second pass average temperature. Nonetheless, the points are reasonably scattered around the 1:1 slope line.

The microstructure being frozen during quenching after the third pass is illustrated in figure(9.23). In this figure, number 4 stands for structure entering the quenching unit.

Figure(9.24) shows the fraction of material which

recrystallized after the third pass was given to the sample
as a function of the third pass average test temperature. At
the same time, the fraction recrystallized for hot rolling
samples is plotted for comparison purposes. Values from
plane strain compression samples are slightly lower than the
ones found for hot rolled samples.

Figure(9.25) shows how $\bar{d}$, the average grain size after the
third pass varies with the average test temperature. Hot
rolling values are also plotted in the same diagram. It can
be seen that both curves run parallel to each other. At
950C, $\bar{d}$ from plane strain compression gives a value 3 μm
higher than the ~27μm obtained for hot rolling samples.

9.5 Hot Rolling Strength Simulation.

Non-isothermal plane strain compression tests were
carried out in order to simulate the hot rolling strengths.
Figure (9.26) shows the dependence of the mean plane strain
strength on the Zener-Hollomon parameter. The full line
represents the results from non-isothermal plane strain
compression tests. The samples were deformed in passes of
25% reduction and the average temperature during the pass
was calculated using Foster's heat flow programme (Foster,
1980). The dotted line represents the dependence of the mean
plane strain strengths as measured from hot rolling samples.
Some results from isothermal plane strain compression tests
are also included in the figure. Clearly, the strength from

non-isothermal plane strain compression tests is higher than the ones obtained for hot rolling or for isothermal plane strain compression. At lower values of Z however ($Z < 10^{17}$), the mean plane strain strength for the non-isothermal tests seems to coincide with the other results. It is not possible from the present set of results to decide wether there is a proportional factor correlating the mean plane strain strength for non-isothermal tests and the other results.

Figure (9.27) shows the dependence of the mean plane strain strength for the second and third passes on the Zener-Hollomon parameter. The full line represents the results from non-isothermal tests whereas the dotted one represents the results from hot rolling samples. There again, the mean plane strain strength is consistently higher than the ones obtained from hot rolling. However, the curves for non-isothermal plane strain compression samples follow a pattern very similar to the one found for hot rolling samples.

9.6 Discussion.

9.6.1   Plane   Strain   Compression   versus   Axisymmetric

Compression Tests.

In the work that has been done so far, it has been suggested that the local hot working variables in a plane strain compression sample may determine the stress-strain behaviour of a material being tested. Axisymmetric compression tests were performed in order to compare their flow curves with the ones from plane strain compression.

It has been previously observed that:

a- The strain to the peak stress, ε_p, in plane strain compression is apparently insensitive to changes in the Zener-Hollomon parameter.

b- Local strains may be responsible for a 'premature' ε_p in plane strain compression tests.

The strain to the peak stress for samples tested under axisymmetric compression is clearly a function of the Zener-Hollomon parameter (figure(9.1)). The absolute ε_p values from axisymmetric compression samples are higher than the ones found for plane strain. Moreover, the dependence of ε_p on Z, through an exponent of 0.15 (equation(9.1)), is considerably higher than the 0.03 exponent found for plane strain compression samples. It is interesting to note that an exponent of 0.15 has been found for C-Mn steels tested mostly in hot torsion (Sellars,1980). However, it has been recently reported (Al-Jouni,1983) that $\varepsilon_p \propto Z^{0.08}$ for austenitic stainless steels tested in torsion.

The strain distribution in a sample being tested in plane strain compression is highly non-uniform, as seen in

figure(7.1). According to this figure, the strain in the diagonal elements or shear bands can be approximately twice the nominal strain applied to the sample. Consequently, local strain rate is also twice the nominal one. Generally, an increase in the strain rate causes the ε_p values to increase. However, as the strain is localized to a small fraction of the total volume of the sample, the increase in temperature may be considerable. An increase in temperature causes the strain to peak stress to decrease. Therefore, ε_p may result from a competition between local increase in the strain rate and local increase in the temperature. At low values of nominal strain rate (say$<5s^{-1}$), the increase in local temperature may not be sufficient to cause a noticeable decrease in the value of ε_p. In fact, the increase in temperature is ~20C for a sample tested at 1025C and deformed to the peak strain (ε~0.35). In this way the nominal strain rate at the low range would cause ε_p to increase. As the strain rate increases further, the rise in the local temperature begins to be significant and ε_p starts to decrease. This phenomenon has been reported for C-Mn steels tested under plane strain compression (Leduc,1980) and it is currently being investigated further in this department (Colás,1983)

The stress-strain curve of 316L tested under plane strain compression is compared with the one from axisymmetric compression in figure(9.3). The stresses for plane strain at $\varepsilon<0.5$ are higher than the corresponding ones for axisymmetric. Thus, the stress at ε 0.1 is ~160MPa for

plane strain as opposed to ~140MPa for axisymmetric compression. The axisymmetric compression sample has to be deformed to $\epsilon \simeq 0.3$ in order to present a stress level of ~160MPa. Therefore, the relation between strains to produce the same stress values from different tests is $\epsilon_{ps}:\epsilon_a \sim 3$ at $\epsilon = 0.1$. On the other hand, $\sigma_{ps} \sim 168$MPa at $\epsilon \sim 0.3$ whereas $\sigma_a \sim 160$MPa. However, $\sigma_a \sim 170$MPa at $\epsilon_a \sim 0.6$. The ratio $\epsilon_{ps}:\epsilon_a \sim 2$ at $\epsilon \sim 0.3$. It is interesting to note that at $\epsilon \sim 0.3$ the ratio is 2, corresponding to the same ratio which is expected between the diagonal strain and the nominal strain in a plane strain compression sample. In addition, ϵ_a needs to be ~0.6, which is the expected strain for the shear band. This suggests that local strains determine the early peak stress in the austenitic stainless steel flow curve obtained from a plane strain compression test. The fact that the ratio $\epsilon_{ps}:\epsilon_a$ is higher at $\epsilon \sim 0.1$ than at $\epsilon \sim 0.3$ suggests that the strain distribution may be more non-uniform at lower strains. This is in agreement with previous findings from gridded aluminium samples tested under plane strain compression (Beynon,1979).

9.6.2 Temperature Evolution During Plane Strain Compression.

Equation(9.3) has been used to describe the dependence of the heat transfer coefficient on the sample surface temperature during air cooling in the Servotest arms. The expression was fed into Foster's two-dimensional heat flow model in order to simulate air cooling in the

Servotest arms.

The apparent emissivity for the sample can be found to be ~0.56 (equation(9.3)). This is lower than the usually reported range of values for emissivity of steels (0.6 to 0.8). The exact value of emissivity depends on the roughness of the sample surface. Plane strain compression samples were machine finished and they usually had an oxide layer when cooling in air. Both facts would cause the emissivity to increase. Therefore, the true emissivity for austenitic stainless steels in the conditions described above ought to be ~0.8. In fact, this is the kind of value found when the sample is cooled in air resting on three nails. In addition, Foster(1981) has reported an apparent emissivity of ~0.57 for Nb-steels air cooled in the Servotest test arms. This value of apparent emissivity is very similar to the one found in the present work. The explanation for this low value of emissivity is given in terms of a 'view factor'. In fact, when the sample is resting on three nails, the upper surface can radiate freely through the upper hemisphere. The bottom surface has less freedom of radiation, however the sample is usually kept at a significant height from the adjacent surface so that the total heat radiated in that direction has negligible effect. In contrast, the sample cooling in air held by the Servotest arms will have a reduced solid-angle over which free radiation from the sample surface can occur causing a lower emissivity coefficient to be measured.

Equation(9.4) describes the surface heat transfer

coefficient of a sample cooling inside the test furnace, held by the Servotest arms. There again, the radiation coefficient is very similar to the one reported by Foster(1981) for Nb-steels (0.414×10^{-10}). Clearly, the radiation coefficient found from equation(9.4) is an apparent one since the emissivity is a physical property and as such will not change with the surroundings where the measurements are being undertaken. Nonetheless, equations(9.3) and (9.4) have been used with reasonable success in the simulation of both air and furnace cooling for plane strain compression samples(figures(9.7) and (9.8)).

9.6.3 Simulation of Hot Rolling Microstructure.

The cooling of a slab being hot rolled has to be reproduced satisfactorily in order to achieve meaningful results during simulation (section(9.4)). It is also important to make sure that the structure entering the first plane strain compression pass is similar to the one from hot rolling slabs. In the present work, most of the rolling was carried out in a 3x25% reduction schedule with 15s interpass time. The same procedure was then used for plane strain compression. Samples were quenched in between plane strain compression passes so that microstructural examination was possible.

Firstly, the structure resulting from a single 25% reduction pass was followed as a function of the first pass

average temperature. Figure(9.12) shows very similar pattern to the one in figure(8.21). In general, the fraction recrystallized for both plane strain and hot rolling is in reasonable agreement. However, the fraction recrystallized from plane strain compression, X_{ps} seemed to be lower than the corresponding rolling one, X_r, for high values of fraction recrystallized (say X>80%). This may have resulted from differences in strain distribution in both samples. Clearly, plane strain compression seems to deform in a more inhomogeneous way than that for rolling samples. Therefore, the time for, say, 95% recrystallized fraction to take place in a plane strain compression sample may be longer than the corresponding one for rolling. If this line of reasoning is correct, then the reverse situation may be observed at low X values (say X<20%). Indeed, figure(9.12) gives X_{ps}~15% at 1050C whereas X_r~10%. Nonetheless, it is acknowledged that the differences involved in X values may not be significant. The average grain size during recrystallization is also plotted against the average temperature during the first pass (figure(9.14)). The overall pattern is very similar to the one obtained for rolling (figure(8.23)) which indicates that the structure after a first hot rolling pass may be simulated by plane strain compression tests. Clearly, given a value of average pass temperature, the fraction recrystallized entering pass number 2 and the average grain size during recrystallization can be interpolated from figures(9.12) and (9.14). Moreover, any first pass rolling can be simulated in this way, provided the average pass

temperature is known. This method of simulation differs markedly from the previous one reported by Foster(1981). He used a plane strain compression sample simulating directly the hot rolling schedule by matching carefully the times and average pass temperatures.

Figure (9.15) and (9.16) show that the results obtained by interpolating from figures(9.12) and (9.14) can reproduce reasonably well the available hot rolling data. Therefore, in principle, plane strain compression tests can be used to follow the recrystallization process occurring during an interpass period in hot rolling.

The mean plane strain strength also seems to be closely simulated by the use of plane strain compression tests as shown in figure(9.17).

The fraction recrystallized and the average grain size show a dependence on the second pass average temperature similar in form to the one found for the structure after the first pass (figure(9.19) and (9.20)). There again, the basic idea was to obtain interpolated values of X and $\bar{d}$ to compare with the ones resulting from hot rolling samples. The idea is schematically represented in figure(9.28). The time interval was always kept constant at ~15s for both plane strain and hot rolling schedules. Additionally, the cooling rate for non-isothermal plane strain compression was the same as for the hot rolling slabs. Therefore, the structure $X_2, \bar{d}_2$ after a pass 2 given at $\bar{T}_2$ can only have been originated from structure $X_1, \bar{d}_1$ given at an average temperature $\bar{T}_1$. The same applies to the

hot rolling structures. Therfore, it is thought that an interpolation procedure can be used as a means of simulation.

Interpolations were carried out in this way, and the values compared with the ones from hot rolling (figures(9.21) and (9.22)). The agreement between measured and predicted fraction recrystallized is reasonable although, for some cases, the fraction recrystallized predicted was ~30% whereas the corresponding rolling one was ~45%. The same procedure was repeated for a third pass. The measured fraction recrystallized and average grain size after the third pass were compared to the ones from hot rolling (figures(9.24) and (9.25)). Clearly, at least 4 different kinds of structure co-exist in a single microstructure after the third pass (figure(9.23)). In addition, discerning between the different structures may not always be obvious. Therefore, there may have been an over-estimation of the values of fraction recrystallized. However, an opposite trend is seen from figure(9.24). In fact, the predicted fraction recrystallized obtained from plane strain compression tests seems to be ~8% lower than the ones from hot rolling samples. Accordingly, the average grain size from plane strain is ~8μm bigger than the hot rolling ones. This may have resulted from an increase in the cooling rate after the third pass, since the sample final thickness was considerably reduced (from ~10 to ~4mm).

9.6.4 Simulation of Hot Rolling Strength.

The mean plane strain strength during non-isothermal plane strain compression tests was considerably higher than the corresponding ones for hot rolling of austenitic stainless steels (see figure (9.26) and (9.27)). The samples were coated with glass lubricant (Dag2626) which melts during the pre-heating period. The tools, at ~230 C, were coated with graphite. When in contact with the tools, the specimen surface cools considerably. As a result, the glass solidifies and lubrication is performed by the graphite only. It may be however, that sticking friction conditions exist in the tool-specimen contact interface. This causes the stress-strain curve of the material to rise, giving a higher strength during the pass. However, the specimens were deformed in passes of 25% reduction each. Therefore, the frictional effects are marginal. Even for a sample which had accumulated the strains of 3 consecutive passes $(\epsilon \approx 1.0)$ since the effects of friction are of significance only at large strains (generally $\epsilon > 1.0$). Precipitate strengthening may also have caused the increase of the mean plane strain strength in relation to the ones obtained for hot rolling. However, the mean plane strain strength, $\bar{\sigma}'$, from isothermal plane strain compression produces reasonable agreement with the ones for hot rolling. This suggests that precipitates may not have accounted for the differences in $\bar{\sigma}'$ obtained for non-isothermal plane strain compression and for hot rolling.

The average temperature during a pass in plane

strain compression has been defined as the arithmetic average between the average temperatures in the cross sections entering and leaving the roll gap. Changing the definition of average pass temperature may have shifted the two curves in figure (6.27) towards agreement. However, at $Z\tilde{}10^{20}$ a decrease in $\bar{T}$ of the order of 150 C is necessary in order to match the values of $\bar{\sigma}'$ from non-isothermal plane strain with the corresponding ones from hot rolling. The magnitude of change in temperature needed suggests that the gradient of temperature inside the sample during test may be an important variable in determining $\bar{\sigma}'$.

In fact, a sample being tested at $\tilde{}800$ C could have its surface in contact with the tools at $\tilde{}400$ C. That is a difference of $\tilde{}400$ C in $\tilde{}8$ to 4mm (depending on the final strain applied to the sample). In these conditions, it seems likely that the specimen being deformed acts as a compound material; the surface layers being cooler and tougher than the center ones.

Compound billets of mild steel with either austenitic or ferritic stainless steels have been extruded (Hughes and Sellars, 1982). These authors have shown that the running pressures for the compound samples are $\tilde{}20\%$ higher than the estimated from solid billets of the component materials. This has been attributed to the increase in redundant work due to the presence of the interface between dissimilar materials.

Slabs of austenitic stainless steel type 316 of 15mm initial thickness have been hot rolled at strain rates varying

between 0.08 to $1.8s^{-1}$ (Aiyedun, 1983). The mean plane strain strengths measured during hot rolling have been reported to be ~30 to 50% higher than the ones calculated from Sims' theory, the difference increasing as the strain rate decreases. Clearly, as the $\dot{\varepsilon}$ decreases the contact time increases. This causes the cooling effect to penetrate deeper into the slab increasing the proportion of tough material to be hot rolled. The same author has reported however, that hot rolling lead at the same strain rates has produced values of mean plane strain strength in agreement with the expected values.

It has also been reported (Castro, 1983) that values of $\bar{\sigma}'$ measured during non-isothermal plane strain compression tests in Al-1%Mn are in reasonable agreement with the ones obtained theoretically. This may be because the aluminium alloy is highly conductive and did not develop the temperature gradient during test as in the case of the austenitic stainless steels.

Figure (9.29) shows the dependence of σ_o on the temperature measured in the centre at the start of the deformation, T^{mc}_{entry} for samples tested in isothermal and non-isothermal plane strain compress The curves for the isothermal and for the non-isothermal tests seem to be parallel. A shift in the temperature axis of ~50 C may bring the curves together. Clearly, the stress-strain curves reach σ_o shortly after the tools touch the sample surface. Therefore, there is little time for a gradient of

temperature to penetrate deep into the sample. As a result the curves from figure (9.29) show a better agreement than the one obtained in figure (9.26) for mean flow stress over a strain of 0.33.

Plotting $\sigma_{0.1}$ as a function of different temperature as specified in figure (9.30) has shown that the average cross section temperature may not be the appropriate independent variable to use. However, when plotting the stress as a function of the average temperature at ~1/5 of the sample thickness away from the tool/sample contact surface, it can be seen that a better agreement between isothermal and non-isothermal tests is obtained, indicating that the chilled surface may have a disproportionately large effect in stresses for deformation in the slip line fields. A similar plot for third passes does not give agreement with the data from rolling, suggesting that some consequential effect possibly precipitation, has been caused by temperature gradients. In view of the magnitude of the discrepancies, the effects of temperature gradients deserve further study.

9.7 Conclusions.

1- The stress-strain curve from a sample tested under plane strain compression differs from the one obtained from an axisymmetric compression test performed under the same conditions. The differences between both curves may be explained in terms of local strain distribution in a plane

strain compression sample.

2- The rate of heat loss per unit area during air cooling of an undeformed plane strain compression sample held in the Servotest arms can be describe by

$$H_s = -0.045 \ (T_s - R_t) - 0.3187 \times 10^{-10} \ (T_s^* - R_t^*)$$

The apparent emissivity is 0.57. This may be explained in terms of the reduced solid angle over which radiation can freely travel when the sample is being held in the Servotest arms.

3- The heat transfer coefficient at the sample surface of a specimen cooling inside the furnace can be reasonably described by

$$H_s = -0.0644 \ (T_s - R_t) - 0.49329 \times 10^{-10} \ (T_s^* - R_t^*)$$

This expression can be used in Foster's two-dimensional heat flow programme to describe the cooling of a sample inside the test furnace during non-isothermal plane strain compression tests.

4- Non-isothermal plane strain compression tests may be used to simulate hot rolling microstructural evolution in austenitic stainless steel type 316L.

5- The mean plane strain strength during non-isothermal plane strain compression tests is higher than the corresponding hot rolling ones. Its thought that the temperature gradient inside the sample being tested may have been an important variable in determining the strength during the non-isothermal tests.

CHAPTER 10

Suggestions for Further Work

1- The computer programme used for hot rolling simulation has been modified since its original version (Leduc,1980) to accommodate several different kinds of materials. This demanded altering the original version in several different parts of its text. It may be of interest to group the lines where modifications are to take place into a single subroutine. In this way, a library of different materials to be simulated can be easily built up in a single programme. This however, may involve significant changes in the logic of the present programme version.

2- A typical simulation of 3x25% reduction pass hot rolling schedule can take as long as 60s of CPU time on an ICL1906S computer. If the programme is to be used for controlling the hot rolling process, then modifications may have to be introduced to speed up the processing time.

The performance of the simplified version may bring some insight into the relative importance of processing and material variables during hot rolling simulation.

3- The present version of the hot rolling simulation programme should be used in an attempt to model industrial hot rolling of austenitic stainless steels in which the situation of partially recrystallized structures in between

stands may arise.

4- The kinetics of static recrystallization and the recrystallized grain size should be studied for the particular case in which the strains applied to the specimen are less than ~0.15.

5- It may be of interest to determine the relationship between kinetics of static recrystallization and strain distribution in plane strain compression tests. The model used in chapter 7 may be further developed to include the effects of local temperature rise.

6- An attempt should be made using non-isothermal axisymmetric compression samples to simulate laboratory hot rolling schedules.

7- The temperature gradient developed during a non-isothermal plane strain compression test should be further investigated so that its effects on the mean plane strain stregth may be better understood.

ACKNOWLEDGEMENTS

REFERENCES

Ahlblom, B., (1977a), Swed. Inst. for Met. Research, IM-1208

Ahlblom, B., (1977b), Swed. Inst. for Met. Research, IM-1210

Ahlblom, B. and Roberts, W., (1978), Proc. Micon '78:
        Optimization of Processing, Properties and
        Service Performance Through Microstructural
        Control, April, 296

Ahlblom, B. and Sandström, R., (1982), Int. Met. Rev., 27,
        nr.1,

Aiyedun, P.O., (1983), Ph.D. research, University of Sheffield.

Alder, J.F. and Phillips, V.A., (1954-55), J. Inst. Met., 83,
        80.

Al-Jouni, F.E., (1983), Ph.D. Thesis, University of Sheffield.

Almasan, C., Nanu, D. and Rosu, I., (1981), Rev. Roum. Sci.
        Tech., Méc. Appl., Tome 26, nr.2, 249.

Amiya, K.C. and Spretnak, J.W., (1975a), Met. Trans. A, 6A,
        733.

Amiya, K.C. and Spretnak, J.W., (1975b), Met. Trans. A, 6A,
        737.

Anderson, W.A. and Mehl, R.F., (1945), Trans. Met. Soc. AIME,
        161, 140.

Assassa, W. and Guiraldeng, P., (1978), Met. Sci., 12, 123.

Avrami, M., J.Chem. Phys., (1939), 7, 1103.; (1940), 8, 212;
        (1941), 9, 177.

Bhargava, R.K., Moteff, J. and Swinderman, R.W., (1976), Met.
        Trans., 7, 879.

Barraclough, D.R., Whiteman, H.J., Nair, K.D. and Sellars, C.M.,
        (1973), J. Testing and Evaluation, 1, 220.

Barraclough, D.R., (1974), Ph.D. Thesis, University of Sheffield.

Barraclough, D.R. and Sellars, C.M., (1974), Mechanichal
        Properties at High Strain Rates, Inst. Phys. (London),
        Conference Series nr.21, 111.

Barraclough, D.R. and Sellars, C.M., Met.Sci., (1979), 13, 257.

Beck, P.A., J. Appl. Phys., (1948), 19, 507.

Beck, P.A., Kremer, J.C., Demer, L.J. and Holzworth, M.L., (1948) Trans. Met. Soc. AIME, 175, 372.

Beese, J.G., (1972), J.I.S.I., 210, 433.

Beynon, J.H., (1979), Ph.D. Thesis, University of Sheffield.

Beynon, J.H. and Sellars, C.M., (1983), unpublished work.

Brown, E.1. and DeArdo, A.J., (1981), Met. Trans. A, 12A, 39.

Bryan, G.F. and Heselton, M.D., (1982), Metals Tech., 9, 469.

Burke, J.E., (1949), Trans. Met. Soc. AIME, 180, 73.

Bywater, K.A. and Gladman,T., (1976), Metals Tech., 3, 358.

Cahn, R.W., (1956), Acta Met., 4, 449.

Cahn, R.W. and Hagel, W.C., (1962), 'Decomposition of Austenite by Diffusional Processes' ed. V.F. Zackay and H.T. Aaronson, Interscience, New York, 131.

Campbell, G.T., Ernest, J.R., Abrahomson, P. and Nicholas,J.G., (1974), Met. Trans., 5, 1875.

Capeletti, T.L., Jackman, L.A. and Childs, W.J., (1972), Met. Trans. ,3, 789.

Castro, F.R., (1983), Private Communication.

Colàs, R., (1982), Research Report, University of Sheffield.

Colàs, R., (1983), Private Communication.

Cole, A.T., (1979), Ph.D. Thesis, University of Sheffield.

Cordea, J.N. and Hook, R.E., (1970), Met. Trans., 1, 111.

Corfield, G.C. and Johson, R.H., (1973), J.I.S.I., 211, 567.

Cotteril P., and Mould, P.R., (1976), 'Recrystallization and Grain Growth in Metals', Surrey University Press, London.

Cuddy, L.J., (1977), 'The Hot Deformation of Austenite' Ed. by J.B. Ballance, Met. Soc. AIME, 169.

Deighton, M., (1970), J.I.S.I., 208, 1012.

Djaic, R.A.P. and Jonas, J.J., (1972), J.I.S.I., 210, 256.

Djaic, R.A.P. and Jonas, J.J., (1973), Met. Trans., 4, 621.

Dutta, B., (1983), Ph.D. research project, University of Sheffield.

El-Kalay, A.K.E.H.A. and Sparling, L.G.M., (1968), J.I.S.I., 206, 152.

English, A.T. and Backofen, W.A., (1964), Trans. TMS-AIME, 230, 396.

Evans, R.W. and Dunstan, G.R., (1971), J. Inst. Metals, 99, 4.

Farag, M.M., Sellars, C.M. and Tegart, W.J. McG., (1968), 'Deformation under Hot Working Conditions', Iron and Steel Inst., Pub.108, 60.

Fisher, J.C. and Fullman, R.L., (1952), Progress in Metal Physics, ed. B. Chalmers, vol.3, 271.

Ford, H., (1948), Proc. Inst. Mech. Eng., 159, 115.

Foster, S.R., (1981), Ph.D. Thesis, University of Sheffield.

Gladman, T. and Wooodhead, J.H., (1960), J.I.S.I., 194, 189.

Gittins, A., Everett, J.R. and Tegart, W.J.McG., (1977), Metals Tech., 4, 377.

Glover, G. and Sellars, C.M., (1972),Met. Trans., 3, 2271.

Guimaraes, A.A. and Jonas, J.J., (1981), Met. Trans. A, 12A, 1655.

Harding, R.A., Ph.D. Thesis, (1976), University of Sheffield.

Hill, R., (1950), 'The Mathematical Theory of Plasticity', Claredon (Oxford), 226 and 254.

Hughes, K.E., (1971), M.Met. Thesis, University of Sheffield.

Hughes, K.E., Nair, K.D. and Sellars, C.M., (1974), Metals Tech., 1, 161.

Hughes, K.E., and Sellars, C.M., (1982), Met. Tech., 9, 446.

Johnson,W.A. and Mehl, R.F., (1939), Trans. Amer. Inst. Min. Met. Eng., 135, 416.

Jonas, J.J., Sellars, C.M. and Tegart, McG., (1969), Met. Rev. 130, 14, 1.

Knudsen, W., Sankar, J., McQueen, H.J., Jonas, J.J. and Hawkins, D.N., (1980), 'Hot Working and Forming Processes' Ed. by C.M. Sellars and G.J. Davies, The Met. Soc., 51.

Korbel, A. and Blaz, L., (1980), Scripta Met., 14, 829.

Kozasu, I., Ouchi, C., Sampei, T. and Okita, T., (1977),
    Microalloying '75, Union Carbide, New York, 120.

Kozasu, I. and Shimizu, T., (1971), Trans. Iron and Steel Inst.
    Jpn., 11, 359.

Larke, E.C., (1957), 'The Rolling of Strip, Sheet and Plate',
    Chapman and Hall Ltd., London.

LeBon, A.B. and Saint-Martin, L.N. de, (1977), 'Microalloying
    '75', Union Carbide, New York, 90.

Leduc, L.A., (1980), Ph.D. Thesis, University of Sheffield.

Leduc, L.A. and Sellars, C.M., (1981), Paper presented at the
    International Conference on Thermomechanical
    Processing of Microalloyed Austenite, Pittsburgh.

Liljestrand, L.G., (1972), Scand. Journal of Met., 1, 271.

Loizou, N. and Sims, R.B., (1953), J. Mech. Phys. Solids, 1,
    234.

Lombry, R., Rossard, C. and Thomas, B.J., (1980), IRSID, RE693.

Luton, M.J. and Sellars, C.M., (1969), Acta Met., 17, 1033.

Martin, P. and Holmes, S.D., (1979), Final Year Report,
    University of Sheffield.

Migaud, B., (1980), 'Hot Working and Forming Processes' Ed.
    by C.M. Sellars and G.J. Davies, The Met. Soc., 67.

Miller, O.O., (1951), Trans. ASM, 43, 260.

Moore, P., (1968), 'Deformation under Hot Working Conditions',
    Iron and Steel Inst., Pub. 108, 103.

Morrison, W.B., (1972), J.I.S.I., 218, 618.

McQueen, H.J., (1968), Journal of Metals, 20, 31.

McQueen, H.J. and Jonas, J.J., (1971), 'Metal Forming
    Interrelation between Theory and Practice',
    Ed. by A.L. Hoffmanner, Met. Soc. AIME, 393.

McQueen, H.J. and Jonas, J.J., (1975), 'Treatise on Material
    Science and Technology. Plastic Deformation of
    Materials, vol.6, Academic Press, New York.

McQueen, H.J., Petkovič, R., Weiss, H. and Hinton, L.G., (1977),
    'The Hot Deformation of Austenite' Ed by J.B.
    Ballance, Met. Soc. AIME, 113.

McQueen, H.J., Wong, W.A. and Jonas, J.J., (1967), Can. J. Phys.,
     45, 1225.

Nörstrom, L.A., (1977), Scand. J. Met., 6, 267.

Petkoviĉ, R.A., Luton, M.J. and Jonas, J.J., (1975), Can. Met.
     Quar., 14, 137.

Petkoviĉ, R.A., Luton, M.J. and Jonas, J.J., (1977), 'The Hot
     Deformation of Austenite' Ed by J.B. Ballance, Met.
     Soc. AIME, 68.

Physical Constants of Some Commercial Steels at Elevetated
     Temperatures (1953), Pub. by Butterworths Scientific
     Publications.

Puchi, E.S., (1983), Ph.D. Thesis, University of Sheffield.

Radu, C.G., Soós, E. and Teodosiu, C., (1979a), Kovové Mat.,
     4, 17, 433.

Radu, C.G., Teodosiu, C., Soós, E. and Dumitrescu, (1979b),
     Met. Research Institute, ICEM, Bucharest.

Radu, C.G., Teodosiu, C., Soós, E. and Dumitrescu, (1981),
     Kovové Mat., 6, 19, 698.

Richardson, G.J., Sellars, C.M. and Tegart, W.J.McG., (1966),
     Acta Met., 14, 1225.

Roberts, W. and Ahlblom, B. (1977), Swed. Inst. Met. Resesarch,
     IM-1209.

Roberts, W. and Ahlblom, B., (1978), Acta Met., 26, 801.

Roberts, W., Bóden, H. and Ahlblom. B., (1979), Met. Sci.,
     13, 195.

Rossard, C. and Blain, P., (1959), Mém. Sci. Rev. Mét.,
     56, 285.

Rossard, C. and Blain, P., (1962), Mém. Sci. Rev. Mét.,
     59, 223.

Ruibal, E., (1983), Ph.D. Thesis, University of San Sebastian.

Ryan, N.D., McQueen, H.J., Jonas, J.J., (1982), ASM, Fall
     Meeting, St. Louis, Mo, USA.

Sah, J.P., Richardson, G.J. and Sellars, C.M., (1969), J. Aust.
     Inst. Metals, 14, 292.

Sah, J.P., Richardson, G.J. and Sellars, C.M., (1973), Indian
     Journal of Tech., 11, 445.

Sah, J.P., Richardson, G.J. and Sellars, C.M., (1974), Met. Sci.,
8, 253.

Sandström, R. and Lagneborg, R., (1975a), Scripta Met., 9,
59.

Sandström, R. and Lagneborg, R., (1975b), Scripta Met., 23,
387.

Sellars, C.M. and Tegart, W.J.McG., (1966), Mém. Sci. Rév. Met.,
63, 731.

Sellars, C.M., (1969), J. Sheffield Univ. Met. Soc., 8, 43.

Sellars, C.M. and Tegart, W.J.McG., (1972), Int. Met. Rev., 17,
1.

Sellars, C.M., (1974), Metalurgia i Odlewarictwo, 63, 297.

Sellars, C.M. and Whiteman, J.A., (1974), Metallurgist, 441.

Sellars, C.M., Sah, J.P., Beynon, J.H and Foster, S.R., (1976),
Report on Research Work supported by SRC grant
nr. B/RG/1481.

Sellars, C.M. and Whiteman, J.A., (1976), Proc. Conf. 'Controlled
Processing of HSLA Steels', York.

Sellars, C.M., (1978), Phil. Trans. Royal Soc. Lond., A288, 147.

Sellars, C.M., (1979), Metalurgia I Odlewnickwo, 5, 377.

Sellars, C.M. and Whiteman, J.A., (1979), Met. Sci., 13, 187.

Sellars, C.M., (1980), Proc. Conf. 'Hot Working and Forming
Processes', Book 264, Met. Soc. Pub., London.

Sellars, C.M., (1981a), Metals Forum, 4, 75.

Sellars, C.M., (1981b), Paper presented at the 24ème Colloque
de Métallurgie, Institut National des Sciences et
Techniques Nucleaires, Saclay.

Sellars, C.M., (1981c), 'Quatitative Metallography', Escuela
Superior de Ingenieros Industriales, San Sebastian,
Spain.

Semiatin, S.L. and Holbrook, J.H., (1983), Met. Trans., 14A,
1681.

Semiatin, S.L. and Lahoti, G.D., (1981a), Met. Trans. A, 12A,
1705.

Semiatin, S.L. and Lahoti, G.D., (1981b), Met. Trans. A, 12A,
1719.

Semiatin, S.L. and Lahoti, G.D., (1983a), Met. Trans. A, 14A, 105.

Semiatin, S.L. and Lahoti, G.D., (1983b), Met. Trans. A, 14A, 743.

Sheppard, T. and Wright, D.S., (1980), Met. Tech., 7, 274.

Silva, P.S.C.P., M.Met. Thesis, (1966), University of Sheffield.

Sims, R.B., (1954), Proc. Inst. Mech. Engrs., 168, 191.

Singer, A.R.E. and Bailey, J.A., (1963-64), J. Inst. Metals, 92, 404.

Soós, E., Teodosiu, C. Nicolae, V. and Radu, C.G., (1981a), Rev. Roum. Sci. Tech., Mèc. Appl., Tome 26, nr.2, 95.

Soós, E., Teodosiu, C. Nicolae, V. and Radu, C.G., (1981b), Rev. Roum. Sci. Tech., Mèc. Appl., Tome 26, nr.2, 179.

Sritharan, T. and Jones, H., (1981), Met. Sci., 15, 365.

Stüwe, H.P. and Ortner, B., (1974), Met. Sci., 8, 161.

Svenson, U., (1977), 'The Hot Deformation of Austenite' Ed. J.B. Ballance, Met. Soc. AIME, 499.

Teodosiu, C., Nicolae, V., Soós, E. and Radu, C.G., (1979), Rev. Roum. Sci. Tech., Mèc. Appl., Tome 24, nr.1, 13.

Towle, D.G. and Gladman, T., (1979), Met. Sci., 13, 246.

Tseng, A.A., (1982), Paper presented at Conf. 'Numerical Methods in Industrial Forming Processes', Swansea, 767.

Tvergaard, V., Needleman, A. and Lo, K.K., (1981), J. Mech. Phys. Sol., 29, nr.2, 115.

Vaugham, T. A., (1968), 'Deformation Under Hot Working Conditions', Iron and Steel Inst., Pub. 108, 68.

Wallquist, G.W., (1962), Jerkontorets Ann., 146, (2), 53. (Trans. BISITS nr. 4028).

Watts, A.B. and Ford, H., (1952-53), Proc. Inst. Mech. Eng., 1B, 448.

Watts, A.B. and Ford, H., (1955), Proc. Inst. Mech. Eng., 169, 1141.

Weiss, I. and Fitzsimons, G.L., Mielityinen-Tütto and DeArdo, A.J., (1982), Paper presented at the International Conference on Thermomechanical Processing of Microalloyed Austenite, Pittsburgh, The Met. Soc. of AIME, 33.

Weiss, I. and Jonas, J.J., (1979), Met. Trans. A, 10A, 831.

Whittaker, H.J., (1973), Ph.D. Thesis, University of Sheffield.

Wong, W.A. and Jonas, J.J., (1968), Trans. Met. Soc. AIME, 242, 2271.

Wray, P.J., (1977), 'The Hot Deformation of Austenite' Ed. by J.B. Ballance, Met. Soc. AIME, 86.

Wusatowski, R., (1966), J.I.S.I., 204, 127.