7.4 The Relationship Between Microstructure and Macrostructure

There is no doubt that macrostructure and microstructure are linked and that a change in one will bring about a change in the other. However it is not entirely certain whether the microstructure is determined by the macrostructure or vice versa. Furthermore the relationship between primary arm spacing and grain size is not known. Previous researches suggested either a directly proportional relationship[243], i.e. as one increases the other one also increases and vice versa or no relationship[316] at all. This section is an attempt to clarify the relationship between the microstructure and macrostructure of as-cast metals utilizing the data obtained for the as-cast structure of AISI310S during the course of this work and the model proposed for primary arm spacing in the previous section.

In order to simplify the picture the following points should be stated:

(i) the change in the width of a single macro grain per unit time (that is in the direction perpendicular to the direction of heat flow) is directly proportional to the change in the primary dendrite arm spacing per unit time. This relationship could be expressed as:

$$\frac{\delta gw}{\delta t} = n\,\frac{\delta \lambda_1}{\delta t} \quad \text{..............................7.32}$$

where gw is the columnar grain width, λ_1 is the primary arm spacing, and t is the dimensionless time which is zero at the onset of nucleation in the melt and n is the number of primary arms in the grain. In strongly columnar structures as in the present case and as in almost all continuous cast structures, this relationship could also be expressed as follows:

$$\frac{\delta gw}{\delta X} = n \frac{\delta \lambda_1}{\delta X} \quad \text{.......... 7.33}$$

where X is the distance between the moving solid/liquid interface and the point where a specific grain is nucleated.

Integrating equations **7.32** and **7.33** would give the same result, assuming the integration constants to be zero:

$$gw = n\lambda_1 \quad \text{.......... 7.34}$$

(ii) However, there arises the question of how the number of primary arms in the grain is determined. If the constitution of the alloy and other casting parameters such as mould size and shape, ambient temperature, amount poured and time of pouring, are kept constant, superheat becomes the only influencing variable. In the present work pouring was carried out at predetermined degrees of superheat and the primary arm spacing might be expected to be directly related to superheat.

Based on the grain width and dendritic arm spacing measurements carried out on ingots cast from different temperatures, and the regression analyses done using these results, a quasi-empirical model will be proposed to formulate a relationship between the grain width and the primary arm spacing. This relationship is linked to the theory of constitutional supercooling. An attempt will be made to derive a model that will fit the results of experiments carried out with superheat as the only controlled variable.

When the data for grain width measured at 6 mm and 14 mm distances were analyzed statistically, the equations found were of the form:

$$lngw = a \quad b(\frac{1}{\Delta T}) \ldots\ldots 7.35$$

This equation can also be expressed as:

$$gw = e^{(a + \frac{b}{\Delta T})} \ldots\ldots 7.36$$

The regression equations gave the relationship between grain width and ΔT for base metal as:

$$lngw_6 = -1.778 - 113.711\ (\frac{1}{\Delta T}) \ldots\ldots 7.37$$

and

$$lngw_{14} = -1.784 - 81.513(\frac{1}{\Delta T}) \ldots\ldots 7.38$$

They could therefore be expressed as:

$$gw_6 = 0.169\ e^{-\frac{113.711}{\Delta T}} \ldots\ldots 7.39$$

and

$$gw_{14} = 0.168\ e^{-\frac{81.513}{\Delta T}} \ldots\ldots 7.40$$

As can be seen, although the point of measurement moved from 6 mm to 14 mm, the coefficient of the right hand side remained almost constant, whereas the numerator of the exponential term dropped by **1.395** times. The significance of these observations will be shown in relation to the results of the analysis of arm spacings since these analyses could explain the difference in the values of the numerators of the exponential terms in equations **7.39** and **7.40**.

When the base metal data for primary arm spacings measured at 6 mm and 14 mm distances were analyzed statistically, the equations found were of the form:

$$\lambda_1 = a + b \ln\frac{1}{\Delta T} \quad\text{.......................................7.41}$$

This equation could be expressed in a general form as proposed in the previous section.

Thus,

$$\lambda_1 = \frac{D}{\alpha_1 R^a} \ln \frac{mC_o(1-K_o)\,\alpha_2 R^b}{DK_o\Delta t} \quad\text{......................7.42}$$

Using the regression equation for λ_1 at 6 mm from the chill zone:

$$\lambda_1\ (6mm) = 0.04733 + 0.007354 \ln\frac{1}{\Delta T} \quad\text{..............7.43}$$

it can be shown that

$$\frac{D}{\alpha_1 R_6^a} = 0.00735 \quad\text{..................................7.44}$$

and,

$$0.04733 = 0.007354 \ln \frac{mC_o(1-K_o)\,\alpha_2 R_6^b}{DK_o} \quad\text{..............7.45}$$

which gives:

$$\frac{mC_o(1-K_o)\,\alpha_2 R_6^b}{DK_o} = 623.87 \quad\text{..........................7.46}$$

If

$$\frac{mC_o(1-K_o)}{K_o} = 1.7 \text{ for } 25\% \text{ Cr in iron,}$$

and $D = 3 \times 10^{-3}\ cm^2s^{-1}$

then

$$\alpha_2 R_6^b = 1.10095 \ldots\ldots 7.47$$

The same argument may be applied to the regression equation for λ_1 at 14 mm from the chill zone:

$$\lambda_1\ (14mm) = 0.09485 + 0.01480 \ln\frac{1}{\Delta T} \ldots\ldots 7.48$$

which gives

$$\frac{D}{\alpha_1 R_{14}^a} = 0.01480 \ldots\ldots 7.49$$

Thus:

$$0.09485 = 0.01480 \ln \frac{mC_o(1-K_o)\,\alpha_2 R_{14}^b}{DK_o} \ldots\ldots 7.50$$

giving,

$$\frac{mC_o(1-K_o)\,\alpha_2 R_{14}^b}{DK_o} = 606.89 \ldots\ldots 7.51$$

and

$$\alpha_2 R_{14}^b = 1.07098 \ldots\ldots 7.52$$

The ratio of the growth term at 6mm to the growth term at 14mm is:

$$\frac{\alpha_2 R_6^b}{\alpha_2 R_{14}^b} = \frac{1.10095}{1.07098}$$

$$= 1.02798 \quad \ldots\ldots\ldots\ldots\ldots\ldots \quad 7.53$$

If **b** is assumed to have the value found from the statistical analysis in the previous section, i.e. **0.08**, then the ratio of growth rates at these two distances becomes:

$$\frac{R_6}{R_{14}} = (1.02798)^{1/0.08}$$

$$= 1.4119 \quad \ldots\ldots\ldots\ldots\ldots\ldots \quad 7.54$$

This suggests that the growth rate drops **1.412** times as the solidification front proceeds from 6 mm to 14 mm from the chill zone

At this stage, the exponential terms in the regression equations for grain width (equations **7.39** and **7.40**) should be reconsidered. It was found that the numerator of the exponential term decreased **1.395** times, as the point of measurement moved from 6 mm to 14 mm. This value is very close to the reduction in growth rate found between the same lines of measurement which was **1.412**. So it can be assumed that the reduction in the numerator of the exponential term arises from the change in the growth rate and the equation for grain width can be reformulated as:

$$gw = \frac{D}{\beta_1} \exp\left[\frac{-mC_o(1-K_o)\beta_2 R}{DK_o \Delta T}\right] \quad \ldots\ldots\ldots\ldots\ldots \quad 7.55$$

As can be seen, the R term in the coefficient on the right hand side of the equation was omitted because, empirically, the value of this coefficient was found to be insensitive to the change in the distance from the edge of the ingot. Even if the R term were included its exponent would be small enough to make its value approach unity.

If the regression equation for λ_1 is considered again, and when equation **7.44** is divided by equation **7.49** the following result is obtained.

$$\frac{R_{14}^{a}}{R_{6}^{a}} = \frac{0.007354}{0.01480}$$

$$= 0.49689 \quad \ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots \quad 7.56$$

It was shown above that

$$R_6 = 1.4119\ R_{14} \quad \ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots 7.54$$

Then by substitution

$$\frac{R_{14}^{a}}{(R_{14} * 1.4119)^{a}} = 0.49689 \ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots 7.57$$

When the equation **7.57** is solved for **a**, it is found that

$$a = 2.0276 \approx 2.0$$

So, the general equation for λ_1 can be rewritten as:

$$\lambda_1 = \frac{D}{\alpha_1 R^2} . \ln \frac{mC_o (1-K_o)\, \alpha_2 R^{0.08}}{D \cdot K_o \cdot \Delta T} \ldots\ldots\ldots\ldots\ldots\ldots 7.58$$

It should be noted that equations **7.35** and **7.58** are sensitive to two main parameters:

(i) superheat

(ii) growth rate

Growth rate is also a function of superheat and distance from the chill zone in this case. However the assumption that, at a given distance from the chill zone, the growth rate stays constant for different amounts of superheat, might introduce a small error in the results when these equations are applied to other alloy systems. As could be noticed, with the data obtained from this work, it was the change in growth rate that was calculated, not the growth rate itself. Therefore, the possibility still exists that these equations could reflect the influence of growth rate on the structure which may change with the degree of superheat at a given distance from the chill zone.

Constants α_1 and α_2 for primary arm spacing, and β_1 and β_2 for grain width are believed to act as correction factors, compensating for the actual growth rate, the thermal gradient and the constitutional term which could not be measured or calculated precisely in this work

However, the strength of these quasi-empirical equations comes from the fact that they have proved to hold over a casting temperature range of 150°C and they reflect the change in macro-grain size and in primary arm spacing with the change in the distance from the chill zone.

These equations suggest that when the constitution of the alloy, the shape, size and mould material as well as the cooling medium are kept constant, the degree of superheat influences the macrostructure and microstructure as follows:

(i) As the superheat is decreased, the macro grain size gets smaller following an inverse exponential curve

(ii) As the superheat is decreased, the primary dendrite arm spacing gets bigger following a logarithmic curve

(iii) At a given superheat the growth of a macro grain is determined by the growth rate of the dendrites making up that specific grain, which becomes a function of the distance of the solid-liquid interface from the chill zone

(iv) At the limit, when the rate of heat extraction is very low (and the growth rate is also low), which is usually the case as $\Delta T \to 0$, the coarsest microstructure is obtained. However this is also the condition for the proportion of the equiaxed zone to approach 100%. Thus there is a price to be paid for obtaining an equiaxed zone by decreasing the amount of superheat. This is in the form of a coarser microstructure. This was also obvious in the microstructures displayed in Figure **6.30** for AISI310S.

(v) At a given degree of superheat, grain width and primary dendrite arm spacing respond differently to changes in the constitution of the alloy. If the constitutional parameter $\left(\frac{mC_o(1-K_o)}{K_o}\right)$ increases, the grain width gets smaller. However the dendritic arm spacing tends to coarsen. Thus if a refined macrostructure is to be obtained by adding constitutional supercooling promoters, a coarser microstructure seems to be unavoidable, as the above equations suggest.

As can be seen from Figures **7.11** and **7.12,** when plotted, the relationship between grain width/primary arm spacing and superheat is very similar to the curve obtained by ploting an equation of the general form:

$$y = \frac{1}{e^{x} \ln x} \quad \text{.......................................7.61}$$

with dimensionless coordinates. The only difference between the two relationships is that the former approaches unity as **x** or **1/superheat** approaches infinity whereas the latter approaches zero.

The physical explanation of the behaviour of the former curve is that at the limit which is 0°C superheat one grain consists of only one single dendrite arm. Equation **7.59** provides the means to support this physical reality: the exponential and logarithmic term on the right hand side of the equation contains the growth rate term **R** which also decreases as the superheat is decreased. The change in **R** counterbalances the change in ΔT to such a degree that equation **7.59** approaches unity as ΔT approaches 0°C.

The similar shapes of the experimental data curve and that of the general mathematical equation stated above strongly suggest that the proposed quasi-empirical relationship (equation **7.59**) can be used to express the relationship between grain width and primary dendrite arm spacing.

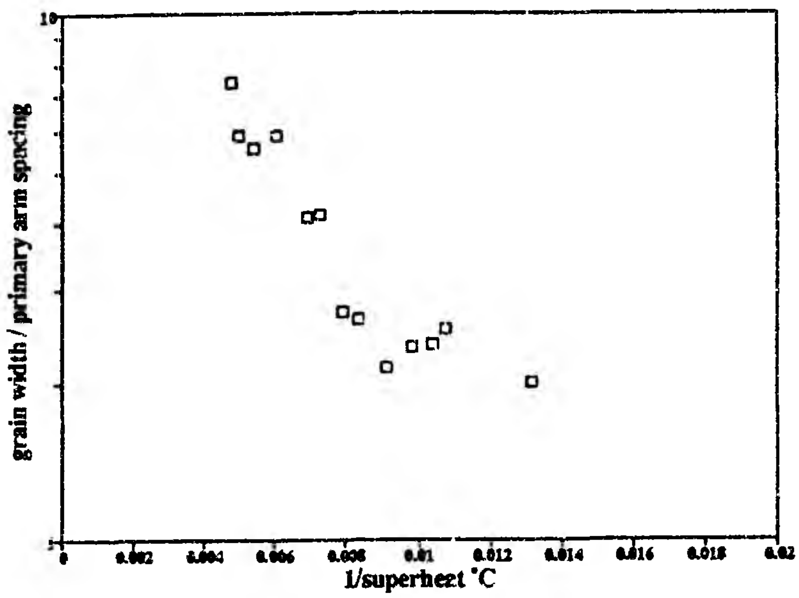

Figure 7.11 The plot of grain width/primary dendrite arm spacing against 1/superheat using the experimental data of this work

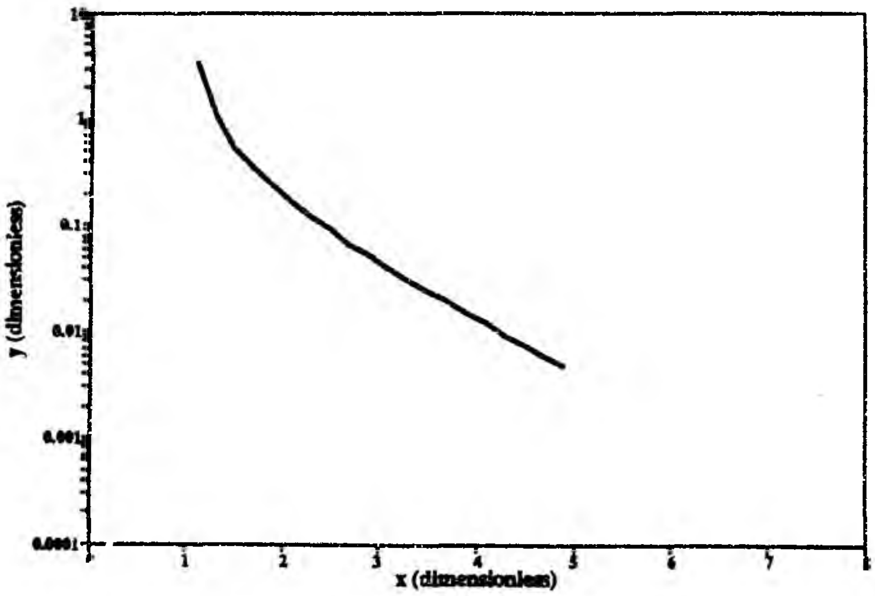

Figure 7.12 The plot of the equation $y = \frac{1}{e^{x} \ln x}$ with dimensionless coordinates

7.5 The Influence of Additions on the Macro- and Microstructure of AISI310S Stainless Steel

The information compiled during the course of this work and the quasi-empirical model proposed for the solidification structure of AISI310S steel enable a clearer understanding as to why the additions made to the melt to refine the macrostructure had no effect on the microstructure. Misch metal and zirconium additions have the potential to refine the macrostructure of the AISI310S steel within a certain casting temperature range. This implies that their potential to promote constitutional supercooling is not big enough to show significant effects where the drop in thermal gradient becomes the main controlling factor in the development of structure. If a stronger constitutional supercooling promoter could be found for this type of steel, it could be effective over a wider temperature range. The same phenomenon was also observed by Fiegenschuh[166] who also studied the effect of cooling rate on the effectiveness of additions and Gautschi[317] who observed that the refining effect was lost when the cooling rate was slower.

It will be appreciated that the major effect of misch metal and zirconium additions was the significant change in growth pattern of the columnar grains of the steel. Misch metal appears to increase the grain width ratio in the casting temperature range 1535-1551°C, due to an increase in the nucleation rate of austenitic grains.

Figure 6.30 clearly indicates that as the grain refinement progresses by the addition of misch metal the microstructure gets coarser. This phenomenon confirms the predictions of the quasi-empirical model suggested in section 7.3.

The sudden drop observed in grain refinement within a very small concentration range of rare earth additions could be related to the fact that, as the concentration of rare earths approaches the solid solubility limit, their activity gets closer to unity, and at the limit it is equal to unity where the addition tends to form compounds with iron minimizing its constitutional effect. A further increase in its concentration provides excess rare-earth which contributes to constitutional supercooling creating the same refining effect again. This phenomenon is typical of cases where the addition promotes constitutional supercooling. The same effect was observed in Fiegenschuh's work[186] where he added rare-earths and misch metal to G-X 15CrNiSi2520 which is constitutionally similar to AISI310S. He related this phenomenon to the grain growth inhibiting effect of the rare earths, i.e. up to a certain concentration of RE, a thin film formed around the grains inhibiting their further growth. However when the concentration reached a certain level it was more favourable for the RE to form a stable compound and therefore, the grain growth inhibiting effect disappeared. As the concentration was increased even further the excess RE reproduced the same effect as it did in the ppm concentration levels. The constitutional supercooling approach and Fiegenschuh's theory may be related as both have the potential to explain the effect. However the relative increase observed in the grain width ratio with addition of misch metal supports the constitutional supercooling mechanism.

This change was more significant in a certain range of residual Ce and La concentration (about 100 ppm). As pointed out in section **7.2**, if the nucleation rate is increased at a given growth rate, the grains start very thin and then get thicker at a rate controlled by the rate of increase of the primary dendrite arm spacing.

This conclusion is in good agreement with the empirical findings of other researchers. Almost all the researchers agree that as the constitutional supercooling or Tarshis parameter increases, the as-cast grain size tends to get smaller, whereas as was mentioned in the previous section, Okamoto et al[314] showed that the primary arm spacing tends to get larger for the alloy systems: Fe-Ni, Fe-Si, Fe-P and Fe-C.

Using equations 7.55 and 7.58, a mathematical expression can now be derived to relate grain width measured at any distance from the chill zone of an ingot cast with a specific superheat to the primary arm spacing measured at the same point as follows:

$$\frac{gw}{\lambda 1} = \frac{\alpha_1 R^2}{\beta_1} \frac{1}{[\exp \frac{mC_o(1-K_o)\beta_2 R}{DK_o \Delta T} \ln \frac{mC_o(1-K_o)\alpha_2 R^{0.08}}{DK_o \Delta T}]} \quad ....7.59$$

simplifying this relationship:

$$gw = \lambda_1 \left(\frac{A}{e^{B/\Delta T} \ln \frac{C}{\Delta T}} \right) \quad ....7.60$$

where $A = \frac{\alpha_1 R^2}{\beta_1}$

$B = \frac{mC_o(1-K_o)\beta_2 R}{DK_o}$ and

$C = \frac{mC_o(1-K_o)\alpha_2 R^{0.08}}{DK_o}$

The relative increase in the grain width ratio with misch metal additions suggests an increase in the nucleation rate of austenitic grains especially in the regions closer to the mould wall, but there seems to be no growth inhibiting process in operation at least in the direction parallel to the mould surface. The structure of the columnar zone obtained in the casting temperature range 1535-1555°C with misch metal addition is similar to that of the base metal cast between 1476 and 1500°C. It should be noted that below those temperatures the grain width ratio of the base metal tends to approach unity again due to the decrease in the growth rate.

As far as the refinement of the macrostructure is concerned, the limit for a single columnar grain is to consist of only one primary dendrite arm. Thus, further refinement in the width of columnar grains does not seem possible once they reach that single dendrite arm size which actually seems to be coarser than those comprising multiple primary dendrite arms. Once they reach that single dendrite arm size, the contribution of additions to constitutional supercooling (in the case of misch metal and zirconium) gets dwarfed by the drastic contribution of a lower degree of superheat, at which the equiaxed zone starts to form with an exponential speed and dominates the structure as the equations suggest.

The effect of zirconium addition in the form of either FeSiZr or pure Zr on grain width ratio is not as significant as that of misch metal additions. However, Zr additions reduced the grain width by about 50% in certain casting temperature ranges, which suggests a grain growth restricting effect is in operation together with an increase in the nucleation rate.

All these observations suggest that refinement of macrostructure via the constitutional supercooling route brings about a coarser microstructure. The secondary arm spacing appears to show a certain refinement in parallel to the refinement of macrograins, but the change remains within the error range of secondary arm measurements. The comparison should be made within a larger range of variation of growth parameters in order to see if a significant trend exists.

7.6 Effect of Macrostructural Changes on the Mechanical Properties of the As-Cast AIS1310S Stainless Steel

The expected, and desired, result of grain refinement would be an improvement in tensile strength or elongation. However this does not occur in all cases. Refining techniques applied to plain carbon and medium alloy steels have produced adequate refinement of the cast grain size, but have failed to result in an improvement in mechanical properties. In fact, these steels have been embrittled by the addition of grain refining inoculants[221]. Other studies on the solidification behaviour of medium alloy steels have indicated that the regions of castings that solidify with a columnar structure have improved strength and ductility compared to portions of the casting with an equiaxed structure.

Flemings[318], and Church et al[319] related the improved properties of columnar structures to the solidification conditions under which they formed rather than the structures themselves. A columnar type of solidification is generally favoured by a steep thermal gradient This steep gradient also provides good feeding conditions for the solidifying metal and results in reduced microshrinkage, and improved strength. This approach also explains the observations of the present work, that as the superheat is decreased the columnar grain size is refined but the strength of the material decreases. However the ingot with an equiaxed zone of 70% showed superior mechanical properties in contrast to observations by previous researchers.

This phenomenon can be related to the following factors:

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This phenomenon can be related to the following factors:

(i) the ingots were cast under argon so there was little likelihood of gases being absorbed in the metal and of macro- or micro-porosity forming in the ingots.

(ii) however in the presence of a small amount of micro-porosity the previous researchers' explanation for better properties with coarser columnar grains needs to be either modified or applied only to sand-castings and/or castings made without a protective atmosphere.

(iii) a possible explanation may stem from the fact that columnar structure is highly anisotropic. However as the columnar grain width gets smaller, so does the degree of anisotropy as shown in Figure **6.50**. There is also a shape effect in that columnar grains have a high aspect ratio as shown in Figure **7.13**. As described in Chapter 6, tensile test pieces were prepared from two orientations perpendicular to each other and to the direction of heat flow. The alignment of the grains in these test pieces is shown schematically in Figure **7.14**.

These schematic Figures explain the difference in tensile test results between samples taken in two directions. In specimen A, the grains tend to align themselves at 45° to the tensile axis direction in order to minimize the shear stresses acting on the test piece. In specimen B this need is minimal, since the alignment of the grains in the test piece is such that the tensile axis and the slip planes (the grain boundaries in this case) are already almost parallel to each other.

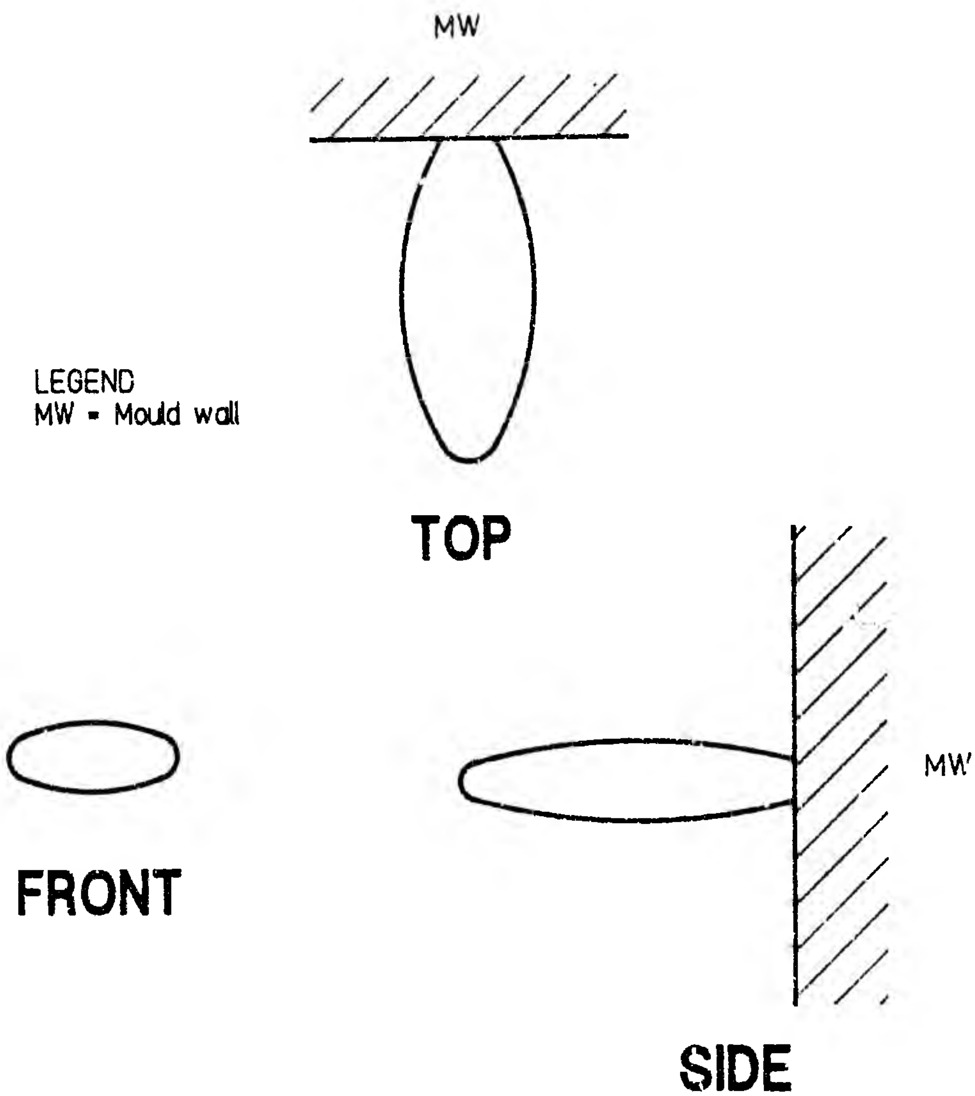

Figure 7.13 Schematic representation of the sections of a columnar grain

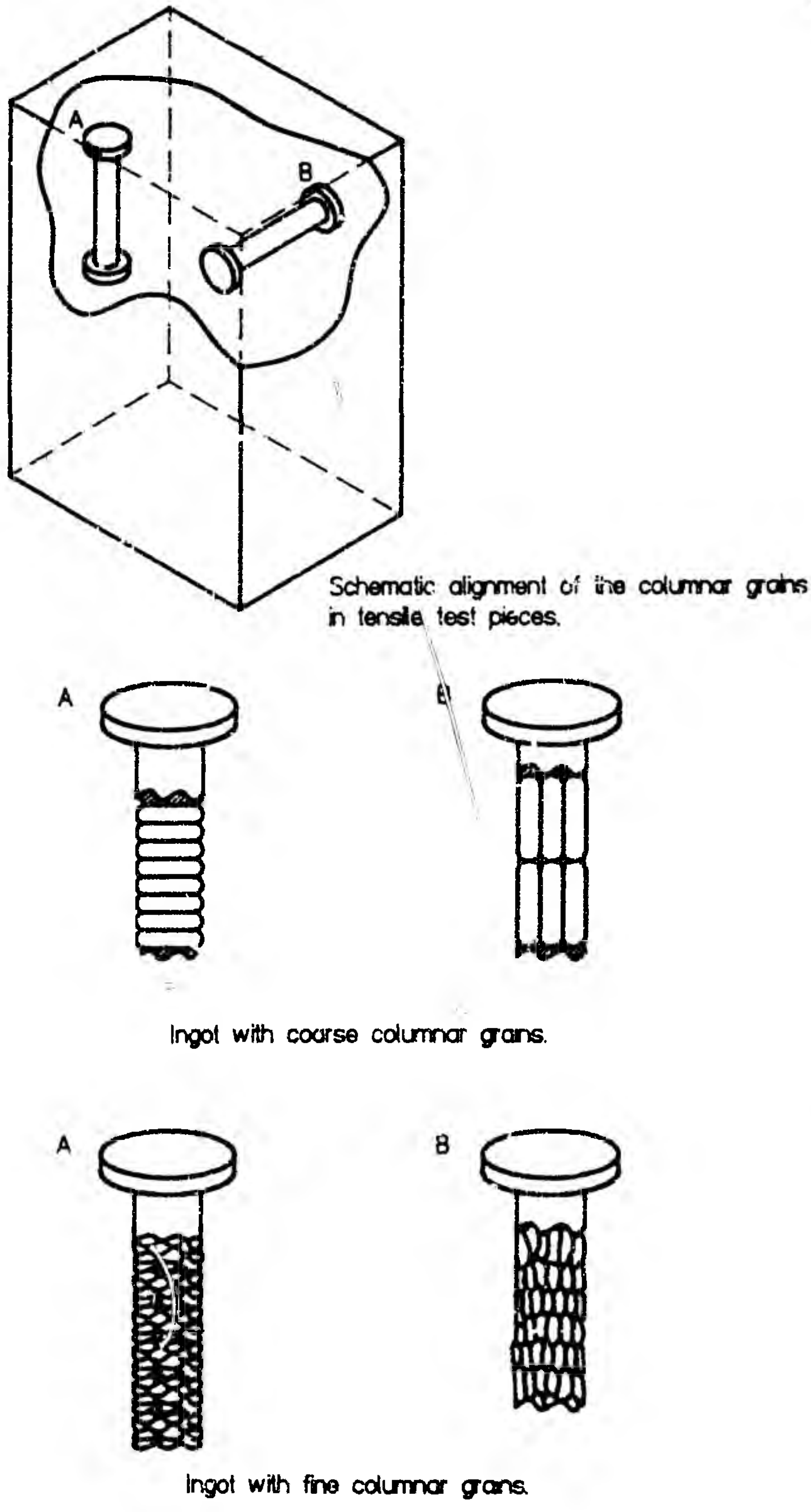

Figure 7.14 **Schematic representation of the alignment of columnar grains in tensile test pieces**

Thus the specimen starts to flow with minimal effort and the tensile strength is relatively low compared to that of specimen A. This difference in the mode of deformation also shows itself in the shape of the fractured surface as shown schematically in Figure **7.15**, and in the photographs of Figure 6.16. Now, as the columnar grain size and the directionality of the grains decrease, this difference in the mode of deformation and fracture also becomes smaller as it was observed in Figures **6.50** and **6.51**. Thus grain size and shape may become the controlling factors in the deformation and fracture mode and in the degree of anisotropy in deformation once the macro- and micro-porosities are brought down to minimal levels. If there is no grain boundary at all (single crystal), then the determining factor would become the strength of the regions between the neighbouring dendrite arms (in the case of castings), i.e. the primary and/or secondary arm spacing play the dominant role in the deformation mode.

This is analogous to a chain with a weak link. In this case the weak link in order of increasing strength may be one of the following:

(i) microporosity
(ii) grain boundaries
(iii) the regions between dendrite arms when only one dendrite is present

In the case of equiaxed structures, a large grain boundary area and an isotropic structure contribute to the higher yield and tensile strengths, and there is almost no difference between the test results obtained in different directions. However, this does not mean that one would always get better strengths with equiaxed structures.

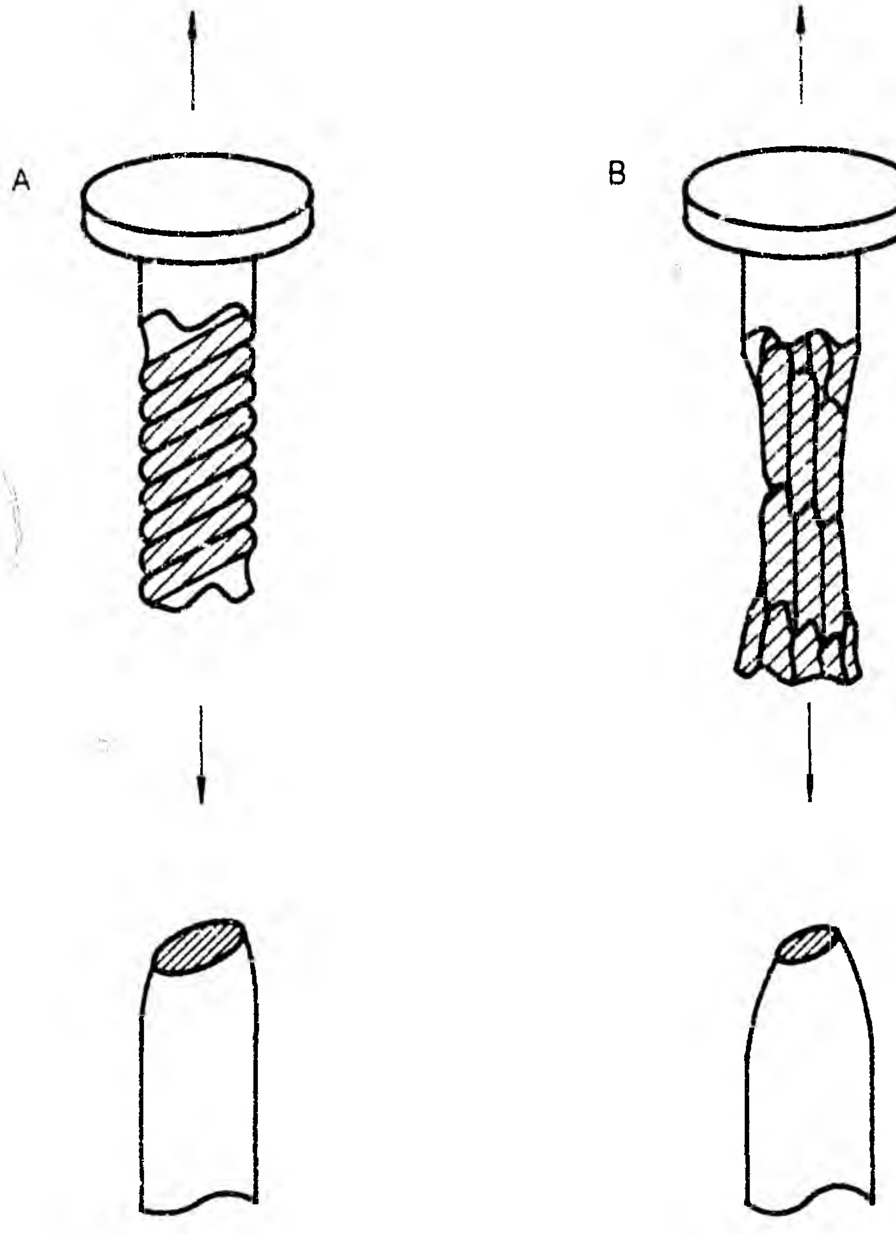

Figure 7.15 Schematic representation of the deformation modes for tensile test pieces A and B and the corresponding fracture surfaces.

If the grain size is coarse as shown in Figure **7.16**, the strengths would most probably deteriorate due to a smaller grain boundary area. The isotropic behaviour of this equiaxed structure would also facilitate the slide of grains over neighbouring grains. They would not be pulled apart as was the case with the coarse columnar structures of tensile specimens cut in direction "A". Thus, the grain size effect is reversed in equiaxed structures. Strength decreases as the grain size is increased, which is opposite to what happens in columnar structures. This approach also explains the results of previous works, obtained with equiaxed materials which gave lower strengths than materials with columnar structures. Their equiaxed structures most probably were not fine enough to give better strength values.

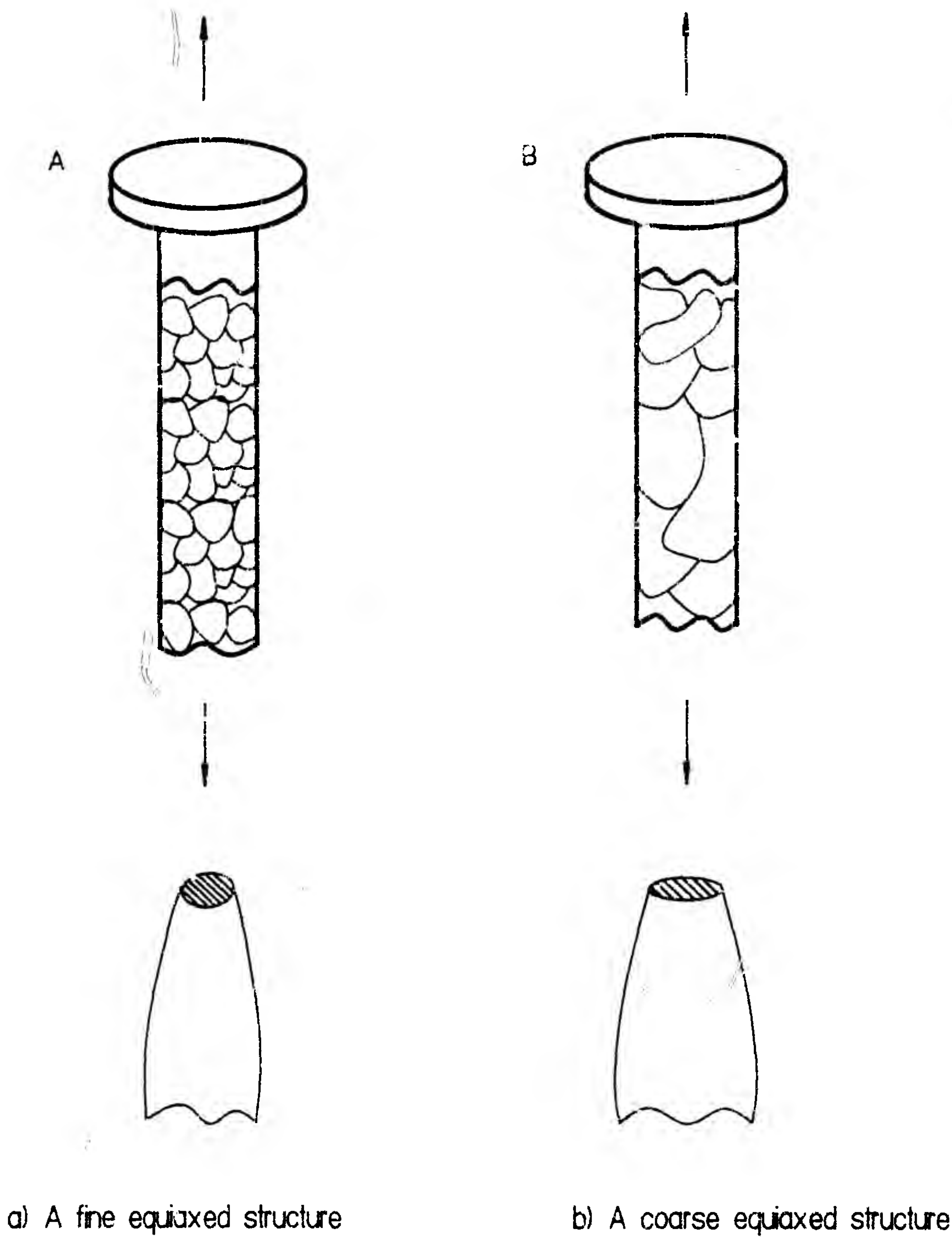

Figure 7.16 **Schematic representations of the deformation mode for tensile test pieces cut from cast ingots with equiaxed structures**

8. CONCLUSIONS

From the experimental results obtained in this study and the quasi-empirical solidification model developed, the following major conclusions can be drawn with regard to the changes in macrostructure and microstructure of AISI310S stainless steel produced under different conditions.

8.1 At cooling rates exceeding a certain value, AISI310S goes through a peritectic reaction during solidification. The cooling rates necessary for peritectic transformation were estimated from the secondary arm spacing of the cast structures. These cooling rates were estimated for the first time to range between about 1400 and 1500°Cs^{-1} . The solidification process may then be expressed as:
liquid $\rightarrow$ δ + liquid $\rightarrow$ γ . Thus under certain conditions AISI310S may transform to delta ferrite first followed by transformation to austenite. This was evidenced by the presence of δ grain boundaries in austenite. At very high casting temperatures primary δ grain boundaries were not observed due to the remelting of the initial δ chill grains and a reduction of the chilling capacity of the mould walls.

8.2 The proportion of the equiaxed zone increased with decreasing alloy superheat. Plots of proportion equiaxed zone against natural logarithm of superheat produced a linear relationship which could be expressed by the equation:

$$\%\ equiaxed\ zone = 244.5 + 49.4 \ln\frac{1}{\Delta T}$$

where ΔT = superheat in °C.

A quasi-empirical equation was developed, which expresses the strong link between the kinetics of formation of the equiaxed zone and constitutional supercooling, in the following form.

$$\%\ equiaxed\ zone = \frac{D}{\alpha R}\ln\left[\frac{mC_o(1-K_o)R}{K_oD\Delta T}\right]$$

8.3 The columnar grain size decreased in an inverse exponential fashion with decreasing superheat. The relationship could be expressed by an equation of the type:

$$gw = e^{(a+\frac{b}{\Delta T})}$$

8.4 Plots of primary dendrite arm spacing against natural logarithm of the reciprocal of superheat were linear indicating that the primary dendrite arm spacing increased logarithmically with decreasing superheat.

The mathematical relationship between these two variables was as follows:

$$\lambda_1 = c + d \ln\frac{1}{\Delta T}$$

It is evident from the mathematical relationships in 8.3 and 8.4 that gw and λ_1 responded differently to superheat. While gw increased with superheat λ_1 decreased.

8.5 The growth rate of a macro grain at a given superheat was determined by the growth rate of the dendrite making up that specific grain. The growth rate of the dendrite was a function of the distance of the solid-liquid interface from the chill zone. The widening rate of columnar grains at different distances from the chill zone was not constant. This phenomenon was described by the term "**grain width ratio**" which was defined as the ratio of grain widths in flat sections at different distances from the chill zone. This ratio was found to approach unity at high and low superheats as a result of the competition between grain nucleation and growth.

8.6 The coarsest microstructure was obtained when the rate of heat extraction was very low, which is usually the case as $\Delta T \rightarrow 0$. However this is also the condition where the proportion of the equiaxed zone approaches 100%. Thus as the proportion of the equiaxed zone increased, due to a decrease in superheat, the primary interarm spacing increased.

8.7 At a given degree of superheat, the grain width and primary dendrite arm spacing responded differently to changes in the constitution of the alloy. When the constitutional parameter (P in equation **4.1**) increased, the grain width decreased. However the dendritic arm spacing tended to coarsen. Thus if a refined macrostructure is to be obtained by adding constitutional supercooling promoters, a coarser microstructure appears to be unavoidable. It has been seen that relatively coarse microstructures were obtained with Misch Metal additions which refined the macrostructures by up to 50%. This effect could be attributed to their ability to provide constitutional supercooling in this type of steel.

8.8 A mathematical expression was derived from the experimental data as follows:

$$gw = \lambda_1 \frac{A}{e^{\frac{B}{\Delta T}} \ln \frac{C}{\Delta T}}$$

where gw is the grain width, λ_1 is primary dendrite arm spacing, and the terms A, B and C contain the growth rate, diffusion rate and constitutional parameters as described in detail in equation **7.47** in Section **7.5**. This equation could describe the changes which took place in the macrostructure and microstructure of the AISI310S stainless steel when cast over a range of 150 degrees of superheat. It strongly suggests that a link between macrostructure and microstructure can be established universally by an equation of this form if the dominating solidification mechanism is constitutional supercooling. However it requires further experimental evidence to develop a general expression which would describe all the alloy systems which solidify under conditions of constitutional supercooling.

8.9 The improved tensile properties obtained in coarse columnar structures and the difference between specimens taken in different orientations could be explained by taking grain size and shape into account. Grain size and shape may become the controlling factors in the deformation and fracture mode and in the degree of anisotropy once the the macroporosity and microporosity are reduced to the minimum. If there is no grain boundary at all (single grain), then the determining factor would become the strength of the regions between the primary dendrite arms. In the case of equiaxed structures, a large grain boundary area and the isotropic structure contribute to the increase in the yield and tensile strengths. There was almost no difference between the test results obtained from specimens taken from equiaxed ingots in two different orientations.

8.10 The shape of the fracture surfaces of the tensile samples revealed that as the grain size decreased, the deformation became more uniform, and it was almost completely isotropic for the ingot with equiaxed structure. Additions of Misch Metal and Zr resulted in more uniform deformation.

8.11 Improved tensile properties were obtained in the as hot rolled condition due to the grain refining effect of the Misch Metal and FeSiZr additions on the steel.

Regression analyses carried out on the results of the hot rolled tensile and impact properties gave the following equations:

$$\sigma_{YS} = 170 + 42.2\ d^{-1/2}$$

$$\sigma_{TS} = 431.2 + 46.1\ d^{-1/2} \quad \text{or}$$

$$\sigma_{TS} = 56.2 - 13.2d + 2.2d^2$$

$$RA = 81.9 - 33.0 + 18.5d^2$$

$$\text{Impact Energy (J)} = 351.3 + 26.6d - 165.75d^2$$

8.12 Cold rolled AISI310S steel with smaller grain sizes in the hot rolled condition showed a smooth edge and no edge cracking while coarse grained material exhibited rough edges and edge cracking upon cold rolling.

The objectives of this study were to examine the influence of chemical grain refining of AISI310S on its macrostructure, its microstructure and its mechanical properties. It is evident that the primary objectives have been met and some useful relationships relating to the solidification, structure and properties of this alloy have been developed.

These studies also revealed an interesting relationship between superheat and structure which was investigated in some detail.

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